Multi-axial failure of high-performance fiber during transverse impact

Matthew C. Hudspeth

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By Matthew Calvin Hudspeth

Entitled
MULTI-AXIAL FAILURE OF HIGH-PERFORMANCE FIBER DURING TRANSVERSE IMPACT

For the degree of Doctor of Philosophy

Is approved by the final examining committee:

Weinong W. Chen                Jeffrey P. Youngblood
Chair                           
James F. Doyle                 Wenbin Yu

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Approved by Major Professor(s): Weinong W. Chen

Approved by: Weinong W. Chen 04/17/2016

Head of the Departmental Graduate Program Date
MULTI-AXIAL FAILURE OF HIGH-PERFORMANCE FIBER DURING TRANSVERSE IMPACT

A Dissertation
Submitted to the Faculty
of
Purdue University
by
Matthew C. Hudspeth

In Partial Fulfillment of the Requirements for the Degree of Doctor of Philosophy

May 2016
Purdue University
West Lafayette, Indiana
To C.P., the one who suffered from my absence
stemming from unnecessary desires.
ACKNOWLEDGMENTS

I must first acknowledge my advisor, Professor Chen, for his unwavering direction and support throughout my graduate studies. I would have not gotten to where I am, nor where I am going without the confidence from such a researcher and leader. Sometimes I would get on my bike in the middle of the night, following the graduate school timeframe, and look up to see Professor Chen still working away to provide support for his numerous students. I will be forever grateful. I would also like to thank Professor Doyle for letting me walk into his office, uninvited and sit down across his desk for hours at a time discussing the mechanics within my topic. I have never met someone so capable, yet so willing, to provide undivided attention to someone of much less stature. One day, I hope I can have such patience. I would also like to thank Professor Yu and Professor Youngblood for taking part in my committee and providing exceedingly strong insight into the my problem when needed.

I don’t think I can ever properly thank the numerous fellow graduate students who have supplied an unheralded degree of time and effort to my various research problems. Too be honest, I have a desire to list every person’s name, but to no avail, my brain will inherently forget someone (probably one of the most important), so out of respect I would prefer to address anyone who has ever provided the slightest bit of assistance; thank you so very much. If ever (and I really do mean ever) you find yourself in need, please do contact me, and I will try my best to return your gracious gestures.

Finally and most importantly, I acknowledge the Lord, Jesus Christ, my savior, for providing me strength to carry through heart wrenching, mundane, and exciting times. His grace abounds.
This body of work is essentially composed of two main topical directions, which separated into two volumes. The first consists of various studies directed at understanding failure in soft body armor, while the second is focused on developing an in-situ characterization technique capable of resolving material damage and failure during high-rate loading. Although these two topics may at first seem rather disparate in nature, effort has been directed at using the secondary topic to shed insight on material failure modes exhibited by fabric armor constituents. That said, the latter topic is much more powerful in nature than to be limited to just fabric armor systems. Indeed, our group has successfully looked at a myriad of different material and structural phenomena, such as, but no limited to: particle-interaction/particle-failure [1], glass fracture [2], composite traction laws Levine:2016, biological ligament failure [3], bone fracture, elastic and plastic metal deformation [4], phase changes in metals [4], and single filament failure [5]. As such, introductory remarks of the second topic will focus on material classes outside of body armor material as initial experiments and the proof-of-concept were directed at such examples as previously mentioned.

Now with regards to the first research thrust, which is of most immediate interest to my PhD study, some topics strayed from the heart of the research direction (seemingly for the better in many creative applications) due to either funding or whimsical fancy, but throughout the research process, a clear underlying need was felt to determine the validity of current understanding or assumptions extant throughout soft armor mechanics. Ultimately, the variety of research directions luckily allowed for work that stumbled upon the aft quoted armor system assumption, predicking armor constituent material always fail in pure longitudinal tension. Although various directions are discussed throughout the current body of work, it is worthwhile to point out that said assumption was determined to be clearly in error, as even sin-
Single filaments can indeed experience a local stress concentration capable of drastically reducing the supposed failure criterion implemented into many (if not all) fabric modeling environments.

Finally, it must be stated that the majority of this work has been reproduced from various journal articles that the author was fortunate enough to write, but much credit must be given to the vast amount of essential collaboration, which without, would have resulted in a loss of a large portion, if not terminating all of this body of work. Experiments have become too detailed and too complex to run entirely on one’s own doing, and as such, the author must continually acknowledge, and fervently praise the efforts given up by fellow collaborators.
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11.3 (a) Diffraction geometry. $O$ is where the X-ray beam and the sample interact; $O'$ is the transmitted beam position on the detector plane; $A$ is the scattered beam position on the detector; $k_i$ and $k_f$ stand for the wavevectors for the incident and outgoing beam, and $q$ is the scattering wavevector. $2\theta$ (i.e. angle $O'O_A$) is the diffraction angle. (b, c) The pixelated scattering vector $q$ map (b) and azimuthal angle $\varphi$ map (c) on the ICCD, when the detector rotation angle is 25°, X-ray energy is 12.9 keV and the beam position is assumed to be vertically centered with the detector.

11.4 (a) Raw voltage signals relevant to the high-rate loading produced from the Kolsky bar setup. Note that in lieu of a transmission bar, a fast-response quartz load cell has been utilized, which results in a reduced experimental framework footprint required by the constrained hutch dimensions. This load response is represented by the dotted black line. Furthermore, this load detection approach is viable if the impedance mismatch between the incident bar end and sample is quite large, which is demonstrated by the similarity between the incident and reflected waveforms shown by the solid black line. (b) Resulting force and strain-rate histories represented by dotted black and solid black curves, respectively. Note that the sample is loaded into the constant strain-rate regime.
11.5 Schematic of the timing sequence used throughout the experimental duration. At time $t_4$ a delay generator (DG) signal is sent to open the gas gun valve, thereby firing the striker. Governed by a specified delay dictated from a predetermined striker travel time, an additional DG signal is sent to open the slow shutter at time $t_3$, which is the opening bracket for the outer experimental window. Note that the slow shutter opening signal is sent early enough to ensure that the shutter window is completely open before the striker impacts the incident bar end. Upon striker impact, a tensile stress wave is sent down the incident bar and, as this wave passes through the bar at time $t_0$, it is detected by a set of strain gauges, which are located 84 cm from the sample interface. Upon detection of the stress wave, a delayed DG signal is sent to close the fast shutter system, demarcated by time $t_1$. Finally, an ultimate DG signal is sent to close the slow shutter system at time $t_3$, thus closing the outer experimental time window.

11.6 Representative stress-strain curve from the 1100-O aluminium samples pulled in tension. The current data, representing a testing strain-rate of $5000 \text{ s}^{-1}$, is demarcated with the solid black curve, and for comparison previous data performed at $1000 \text{ s}^{-1}$ [198,199] have also been included. The various symbols which are overlaid on the plot dictate stress-strain states at which a diffraction pattern has been recorded with the ICCD. In-depth analysis from only an unloaded sample and a sample loaded up to demarcation 10 have been included in the in-depth diffraction pattern analysis shown in Figure 11.8, as the current goal of this study is to show the possibility of performing such high-strain-rate loading while simultaneously capturing high-frame-rate diffraction patterns and phase-contrast images.

11.7 Experimental diffraction patterns wherein the ICCD gating time was varied so as to allow for (a) single-pulse diffraction and (b) 65-pulse diffraction. Azimuthal integration of each pattern was performed to display the corresponding one-dimensional intensity plots for (c) single-pulse diffraction and (d) 65-pulse diffraction. Clearly the multi-pulse gating yields a much higher signal-to-noise ratio, but note that there is still enough information in the single-pulse image to detect a distinct ring pattern and well defined peak position via post-processing.
11.8 Diffraction patterns from the Al sample, collected (a) before and (b) 30 ms after the start of tensile pulling. (c) Diffraction intensity and corresponding simulations from unstrained and strained Al samples. The one-dimensional data profiles were obtained by radially averaging the two-dimensional diffraction patterns shown in (a, b) from azimuthal angle 173° to 187°, as indicated. (d) A closer look at the (111) peaks, showing the shift of the peak due to a lattice strain of 0.25%. Note the disparity in elastic strain of the (111) peak as compared with the large amount of average plastic tensile strain demonstrated in Figure 11.6.

11.9 Representative stress-strain curve produced from the equiatomic NiTi superelastic pulled in tension. Tests were performed at 1000 s⁻¹ and are transformation demonstrated by the solid black curve, which is compared with the dotted black line, representing previous data performed at 1200 s⁻¹ [202]. Similar to the aluminium tests, symbols have been overlaid on the plot demarcating stress-strain states at which diffraction patterns have been captured with the ICCD.

11.10 Diffraction patterns from the NiTi sample, collected (a) before and (b) 1.75 ms after the start of tensile pulling. (c) One-dimensional diffraction intensity profiles, obtained by radially averaging the two-dimensional patterns with all available azimuthal angles. Reference peak positions of the austenite (solid lines) and martensite (dashed lines) phases are displayed at the bottom, and the blue, red, green and cyan colors represent peaks that correspond to the second, third, fourth and fifth harmonic energies, respectively. Note that the photons with the first-harmonic energy have been entirely absorbed by the Si filter.

11.11 PCI sequence captured with the high-speed camera showing the deformation process of the NiTi material throughout the loading process. Frame times have been chosen to correlate with the stress-strain states at which diffraction patterns displayed in Figure 11.12 were captured. Note that this image sequence consists of snapshots within one loading event.

11.12 Diffraction pattern sequence captured with the ICCD which shows the evolution of the martensitic transformation at sequential stress states during the dynamic loading process. It is important to note that each pattern is captured from a different sample within the tensile loading history at stress-strain states represented by the appropriate symbols in Figure 11.9 and at delay times shown in Figure 11.11. The pattern taken at t = 1.75 ms was captured within the PCI image sequence shown in Figure 11.10.

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ABSTRACT

Hudspeth, Matthew C. Ph.D., Purdue University, May 2016. Multi-Axial Failure of High-Performance Fiber During Transverse Impact. Major Professor: Weinong W. Chen.

The effect of projectile nose geometry on ensuing wave development in high-performance yarns is explored during single yarn transverse impact. Special attention has been placed on visualizing the immediate region around the projectile-yarn contact site for 0.30-cal round, 0.30-cal fragment simulation projectiles (FSP), and razor blades using high-speed imaging. Kevlar® KM2, Dyneema® SK76 and AuTx have been impacted at velocities ranging from \(\sim100\) m/s to \(\sim1200\) m/s depending on projectile nose shape, with an emphasis set on determining the critical velocity wherein below said velocity significant development of wave propagation occurs and above said velocity the yarn fails immediately upon impact. In actuality, a range rather than a stark jump defines this critical velocity and as such, ranges are determined for all three indenters yielding increasing values when using the razor blade, FSP, and round projectile heads, accordingly. Fracture surfaces have been analyzed for impact conditions surrounding both the upper and lower ends of the transition regions so as to determine local or long range fiber failure being defined by shearing or fibrillation, respectively. Additionally, above said critical velocities, strikes with 18-mm flat discs and 18-mm round projectiles have been performed in efforts to ascertain the location of yarn failure at the initial stages of impact using high-speed imaging. While failure is seen to occur at the corners of the flat disc projectiles, round indenters yield yarn rupture directly in front of the impacter in a myriad of locations. Additionally, in efforts to determine the longitudinal elastic modulus for both Kevlar® KM2 and Dyneema® SK76, in-situ measurements of longitudinal wave speeds are made during the majority of the transverse impact experiments. Measured wave speeds remain...
unchanged in the range of velocities tested for Kevlar® KM2 and increase slightly with increasing impact velocities for Dyneema® SK76. As expected, the longitudinal wave speed is determined to be independent of projectile nose shape for both fiber types. Finally, dynamic experiments for Kevlar® KM2 are compared to quasi-static results from single KM2 and SK76 filaments loaded in a transverse deflection environment via FSP indenter, round indenter, and razor blade. Single filaments are loaded at angles comparable to those predicted to develop behind the transverse wave front when yarn is impacted at the experimentally determined critical velocity. Good strain correlation is found between said quasi-static experiments and aforementioned transverse impact experiments when using all three projectiles. Additionally, imaged fracture surfaces for low angle and high angle failure are very similar to those found from low-speed and high-speed transverse impact, respectively. Insight of the local stresses developed during transverse impact is also gained through analytical and finite elements modeling of both static and dynamic conditions, with a focus placed on understanding the stresses present within a single filament around the projectile head in both rate regimes. Ultimately results for analytical modelling, computational modelling, quasi-static loading, and transverse impact are compared, yielding good correlation between modelling efforts and experimental results.
1. Introduction

High-performance polymeric fiber is used in a plethora of applications due to its unmatched stiffness-to-weight and strength-to-weight ratios. While industrial applications and sporting equipment such as fishing line, climbing ropes, water vessels, and tires are areas of interest, the direction of the current work is entirely focused on ballistic functions such as body armor and turbine fragment containment systems. Due to the inherent life-saving nature of these applications, it is clearly of great concern for researchers and designers to understand the mechanism by which said materials halt incoming projectiles. Below the ballistic limit of a fabric system, energy dissipation mechanisms chiefly exhibited by principal yarns have been shown to play the major role in halting the incoming projectile, being defined by longitudinal strain energy, longitudinal kinetic energy, and transverse kinetic energy [6]. These energy mechanisms have been shown to be governed by effects such as, but not limited to, fiber-fiber friction, yarn-yarn interaction (sliding, trellising, pull-out), weave structure, environmental aging, projectile nose geometry, and projectile strike angle [7,8]. In contrast, above the ballistic limit of a fabric system, various authors have shown that the effect from many of these mechanisms are attenuated, and failure becomes localized or ‘inelastic’ [7,9–11]. Such an immediate local failure process inherently prevents energy from moving away from the impact site, as both longitudinal and transverse stress wave propagation is undeveloped due to early or premature impact site failure, yielding the consequence of an almost negligible projectile deceleration [12,13].

This stark transition in energy dissipation from full fabric panel involvement at relatively low impacting velocities to rapid local failure exhibited at relatively high impact velocities is not only governed by projectile velocity, but is also coupled to projectile nose geometry, which directly affects both the mode of penetration and the constituent yarn failure. Montgomery et al. [14] and Tan et al. [15] showed that
increasing nose sharpness of impacting projectiles into aramid fabrics allowed for more pronounced fabric windowing/nosing, thereby decreasing the ballistic limit of a fabric system when impacted by increasingly sharp-nosed projectiles. Additionally, Tan et al. [15] demonstrated that flat nosed projectiles were more adept at cutting through the constituent yarns while spherical head impact resulted in longitudinal tensile failure of the yarns beneath the fabric footprint. The sharpness of a projectile nose geometry also controls the ballistic limit of a fabric system, indeed, Abbott demonstrated that both Nylon 6/6 composites and an undisclosed dry aramid fabric revealed $V_{50}$ values coupled to the chamfer angle of both conical nosed projectiles and FSPs [16]. It was alluded that the reduction in armor $V_{50}$ due to sharpened conical projectiles resulted from increased windowing through the fabric weave, while the decrease in fabric $V_{50}$ due to increasing FSP chamfer angle was described as ‘expected’, and reference to Prosser’s [9] explanation of local shearing was mentioned. The cutting effect by sharp-cornered projectiles has also been described by [7] to be driven by local shearing of the constituent yarns, wherein he notes a clear difference is seen when impacting a variety of fabric types/thicknesses with chisel-nosed FSPs and right circular cylinders (RCCs). Interesting to note is the effect of nose geometry on extremely overmatched fabric systems; Gibbon et al. (2014) impacted varying thicknesses of Twaron fabric packs with a variety of different projectile nose geometries at a constant overmatched velocity, and although stated otherwise in report, found nearly similar residual velocities for all projectile types. Additionally, both the number of broken yarns found on the impact surface ply and the full pack energy absorption were seemingly unchanged for all pack thicknesses (1-5 layers). Results from the immediately aforementioned studies [7, 9, 14–17] thus suggest that the response of the fabric system is governed by both structural artifacts and constituent yarn mechanical behavior.

In this light, it is of interest to separate the fabric’s structural and material behavior in efforts to gain any possible understanding of fabric failure in a fashion as simple as feasibly possible. It is thus proposed in the immediate work to clear away the struc-
urally governed energy dissipation mechanisms which can occur both below and near the ballistic limit of a fabric system and rather, focus on the local failure of the constituent yarns and fibers. Rather than moving directly into the most profound aspect of this study, an arial overview of the various progressions which were undertaken by the author will be presented, which ultimately culminated in a reasonable step in the direction towards understanding filament/yarn behavior and failure during transverse impact. A brief background will be given into current understanding of failure of the two main high-performance fiber materials, namely aramids and ultra-high molecular weight polyethylene. Additional effort will be directed at the behavior of filaments when loaded in a multi-axial stress state and examples of typical micrographs found from filament materials when loaded in tension will be given; both topics are initially discussed to provide ancillary background of the various topic discussed throughout this document. It must be noted that only slight introductory remarks are to be made in this opening verbiage, and more detailed and specific introductions are left to each respective ensuing chapter, as it seems better suited to introduce each chapter with it’s respective goal due to the compilation nature of this document.

1.1 Major Types of High Performance Fibers

Molecules composed of amide groups or aromatic structures are typically very stiff and strong; thus the majority of high performance fibers are composed of these sorts of molecular geometries. As such, high-performance filaments traditionally possess a high degree of crystallinity, generally upwards of 90%, resulting in high longitudinal strength and stiffness. Due to said profound strength-to-weight and stiffness-to weight ratios, high performance filaments can be found in numerous applications within automotive, aerospace, and armor protection environments. Currently, there exists two main commercially available high-performance filament materials, namely ultra-high molecular weight polyethylene (UHMWPE) and rigid rod poly(p-phenylene terephthalamide), also known as aramids. Both provide excellent mechanical properties,
and each have slightly varying application realms due to inherent physical properties, which will be expounded upon shortly.

1.1.1 Rigid Rod Fiber (aramid)

The first high-performance material type, which has been commercially available for roughly 50 years, is known as the rigid rod fiber, or aramid. These filament materials are composed of a rigid molecular backbone, incorporating at least aromatic ring [18]. When processed in a fashion to align these molecules in a repeating geometric profile, said aromatic structures create an para-aramid profile, exhibiting a highly aligned molecular structure, which is characterized by a microstructure exhibiting these highly aligned molecules along the longitudinal direction of the fiber [19]. An example of this structure can be seen in Figure 1.1(a), wherein the aromatic structures form into zig-zag sheets that are radially oriented to form the filament, which can be seen in Figure 1.1(b). The transverse forces holding the molecules together is promoted by hydrogen bonding or Van der Waals attraction [20]. Such a stacking sequence allows for extremely high levels of crystallinity, typically in the range of 90-95% [20], and allows for final filament materials that are quite resistant to mechanical, thermal and chemical environments [21].

Major drawbacks of this material type include the complexity of the initial processing and environmental aging effects. The former, initial processing, requires extremely acidic solutions in order to disentangle the rigid rod molecules, therefore allowing molecular alignment via some sort of mechanical processing procedure (e.g. wet jet spinning) [21]. Such mechanical processing procedures are inherently required for aramid fibers, as the oxidative degradation temperature is quite often below the melting temperature of the material, thus forcing the use of highly acidic solutions to spin the molecules to alignment. Although requiring a vigorously controlled production process, along with great effort to properly dispose of manufacturing byproducts, aramid materials are used highly throughout numerous high-performance industries,
with manufactures such as DuPont™ and Teijin™ producing the majority of aramid filaments now seen in service.

The possibility of producing aramid material was first discovered by the late DuPont™ scientist Stephanie Kwolek [23,24], and via patents, resulted in DuPont™ being the only company producing aramid fibers until 1985. Although still the best known high-performance fiber by the public eye, Kevlar® has had a healthy market competition, most notably from the filament material Twaron®. Initially developed by Akzo, this rigid rod molecule filament is now produced by Teijin™, and exhibits very similar mechanical, thermal, and chemical properties to that of Kevlar®, and has found itself in many applications initially held by Kevlar®, especially throughout Europe. Although other material types have sprung up over the last few decades, most notably polyethylene based fibers, which will be discussed shortly, aramids are still the most highly used filament material and find themselves consistently being implemented into high-performance and safety applications.

1.1.2 Flexible Coil Fiber

As just mentioned, the main combatant to aramid materials is ultra-high molecular weight polyethylene (UHMWPE), which is a type of flexible coil fiber, aptly named for the long flexible polyethylene molecular chains [25–27]. Similar to aramids,
UHMWPE filaments possess extraordinary specific stiffness and specific strength, resulting from the high molecular weight chains that are nearly uniform in directionality along the filament longitudinal direction [28]. Said chain directionality is derived from two main processing methods for this material class, namely: (1) flow conditions during the spinning process that promote highly extensional molecular orientation characteristics, and (2) extreme drawing conditions that further align the flexible molecules along the fiber length [18]. An example of a polyethylene mer unit, which is built up into a zig-zag confirmation, can be seen in Figure 1.2(a). Within the actual filament material, UHMWPE is composed of two main microstructural units, namely crystalline and amorphous regions. The crystalline regions, which compose the majority of the filament mass, are typically 60-400 nm in length [28] and provide the longitudinal stiffness of the fiber. In efforts to increase the degree of crystallinity, the PE molecules must be dissolved in some sort of solvent and then drawn to a 60:1 ratio in efforts to reduce the entanglement exhibited by the flexible molecules, thereby allowing for increase chain orientation [18]. Although early versions of UHMWPE fibers had crystallinity values of ∼75%, more recent filament materials have reportedly been able to achieve crystallinity values north of 90%, resulting in filament properties equating to or even exceeding those of aramid materials. As such, several manufacturers produce UHMWPE filament material, most notably DSM™ in the form of Dyneema®, and Honeywell™ in the form of Spectra®.

At this point it is important to note that great effort is taken increase molecular weight while maintaining an adequate level of crystallinity. Higher molecular weights are much more difficult to disentangle during manufacturing, but lead to higher filament strengths. Such increases in strength are due to increases in overall Van der Waals attraction between two chains. As it has been proposed that these materials do not fail via chain scissioning, but rather due to chain slippage, it is suggested that longer chains increase the relative tensile stress state needed to rupture, although there is a maximum chain length wherein further extension will not feel any appreciable mechanical engagement due to the presence of dislocations causing interaction
Figure 1.2. (a) Polyethylene molecule. (b) UHMWPE microstructure showing both crystalline and amorphous regions [28].
load attenuation made previously between the highly extended chain lengths [29]. Furthermore, it is of interest to understand how these filaments do indeed rupture, and if there exists any sort of alternative stress state that is capable of reducing the oft assumed pure tensile failure independent of loading geometry.

1.2 Shear Testing of Single Fiber

In order to better understand the behavior of high-performance fibers compiled into a composite matrix or when woven into a fabric, it is imperative to not only understand the axial properties of the fibers but to also determine their shear modulus as well. Deteresa et al. modified a pre-existing technique originally used to determine the gravitational constant felt throughout the earth [30], in efforts to develop a device which could adequately determine the torsional modulus of a single fiber [31]. This method, called a torsion balance and depicted in Figure 1.3(a), ascertained the torsional modulus of Kevlar 49 to be 1.8 GPa. Also in this study, an analysis was performed in order to better understand the effects of extremely high torsional strains and its respective recoverability in a single Kevlar 49 fiber. It was concluded that applied torsional strains up to ∼10% are in effect elastic, and beyond this amount there are permanent depreciable effects on the recovered strain in the material. This damage characteristic, which was analyzed quite extensively, can be seen in Figure 1.3(b).

This torsional pendulum device was further extended by Mehta and Kumar in efforts to better understand the damping effects present in the experimental device resulting from friction effects of standard atmosphere [32]. They placed their device into a vacuum chamber and found that at lowered pressures, the torsion modulus of the fiber increased, due to the reduced damping factor emanating from air friction effects during rotational motion. In this study, it was found that the torsion modulus of Kevlar 49 was 1.4 GPa.
Figure 1.3. (a) Schematic of torsional pendulum used to test shear moduli. (b) Recoverable torsional strain vs. applied torsional strain in Kevlar 49 fibers [31].

Figure 1.4. (a) Schematic of torsional pendulum system inside of a vacuum chamber. (b) Effect of damping on shear modulus and damping factor [32].
1.3 Transverse Testing of Single Fibers

While high-performance fibers possess a very steep modulus and elevated tensile strength in the axial direction, it has been shown they reflect a great deal of anisotropy in mechanical properties when loaded in the transverse direction. A method of determining the transverse modulus of single fibers was developed by a research group from I.C.I. Fibres Limited, wherein a single fiber was placed between two parallel plates and loaded transversely along its length [33, 34]. This method can be seen in Figure 1.3. Lateral compression tests performed by Cheng et al. found that the transverse compressive and longitudinal tensile elastic moduli were 1.34 GPa and 84.62 GPa, respectively, clearly depicting a highly anisotropic microstructure [35]. This sort of result can be explained by the highly crystalline microstructure running along the direction of the fiber axis as reviewed in section 1.1. Along the axial direction of the fiber, there exists highly covalent bonding along the chain length, while in the transverse direction, there exists mostly secondary Van der Waals and hydrogen bonding, thus producing a very weak transverse fiber. This same sort of analysis was performed by Lim on the high performance A265 fiber resulting in a transverse compression Young’s modulus of 1.83 GPa and an axial tension Young’s modulus of 159 GPa, once again clearly depicting a highly anisotropic material [36]. Clearly these filament materials are affected by the actual loading state to which they are subjected.

1.4 Common Failure Mechanisms in Fibers

As micrographs depicting the rupture surfaces of fibers subjected to various loading conditions throughout this document, it is deemed useful to provide a basic background into the modes of failure that are exhibited during failure of single filaments, specifically when pulled in tension.

It is well known that failure of a fabric composed of high-performance fibers is a function of the properties of the penetrating projectile along with the inherent fabric
Figure 1.5. (a) Experimental apparatus used to compress single fiber transversely and measure the contact width. (b) Schematic of single fiber between two rigid platens with no transverse strain. (c) Contact zone, width 2b, of single fiber on rigid platen when placed in a transversely strained condition. [33,34].
properties such as areal density, number of fabric plies, fabric type, friction between the yarns and fibers themselves, and friction between the yarns/fibers and projectile [7]. Of current interest to this study is the failure of the fibers themselves and the relevant modes of deformation and fracture. In the two most common aforementioned fiber types, Aramid and UHMWPE, the following failure mechanisms are most prevalent: (1) fibrillated break, (2) pointed break, (3) kink band break, (4) crazing, and (5) plate formation. These are described in the following:

1. Fibrillation: The most common failure mode exhibited during tensile failure of aramid filaments. It is is a direct result of fibrillar structure seen within the filament and due to the poor level of transverse bonding. Such a failure mechanism can be likened to the failure of a multi-braided rope. An example of the fibrillation behavior can be seen in Figures 1.6(a) and 1.7(a) for an aramid and UHMWPE, respectively.

2. Pointed break: A subset of fibrillation, pointed break occurs when fracture results in just one fibril, rather than the more common multiple fibril failure zone. Typically a point break results in two similar fracture surfaces from the mating ends, and upon first glance, mimics a necking phenomenon, though necking is most definitely not the means of failure. As with fibrillation, fracture is still quite brittle and is believed to occur via breakage of the weak Van der Waals or hydrogen bonding in the filament transverse direction. Pointed break is also often called axial splitting [37].

3. Kink band break: The least common method of failure, kink band breakage occurs in a very localized fashion, and as the name implies, at a kink band present in the filament. Typically such a failure requires a fatigue loading condition, either via bending or axial loading [38], which promotes local damage either by dislocation pile-up along the molecular chains or from voids that could form between the skin-core structure [37]. An example of a kink band break can be found in Figure 1.6(c).
Figure 1.6. Common failure modes seen from aramid filaments when loaded in tension: (a) fibrillation, (b) pointed break / axial splitting, and (c) kink band break [39].

Figure 1.7. Common failure modes seen from UHMWPE filaments when loaded in tension: (a) fibrillation and (b) plate fracture.

4. Plate formation: This mode of failure is also quite uncommon, and is believed to occur due to failure within the crystalline regions of the material, thereby allowing for a localized slip. An example of this failure process can be seen in Figure 1.7(b).

1.5 Cunniff Dimensionless Parameter

As a bit of an aside, one the most influential analysis tools currently employed during the design and screening of a high performance fibers is a set of dimensionless parameters recently developed by Cunniff [10]. This optimization technique, which
very adequately represents the vast majority of high-performance fibers currently in use, allows for a better understanding of the relevant mechanical properties of a specific fiber and how these extensive and intensive properties effect the ballistic performance of an overall fabric ply or composite. In order to dimensionalize the system, relevant constituent properties of both the fiber incorporated into the fabric ply and the projectile were determined. First, regarding the projectile, it is assumed the only relevant mechanical properties are mass and face area. Second, it is proposed that two characteristic parameters can describe the armor package: fiber specific toughness and fiber strain wave velocity. Importantly, it is assumed constituent fibers follow a linearly elastic stress-strain behavior, rendering toughness as a product of fiber failure stress and failure strain, which is calculated from quasi-static yarn tensile tests. This assumption has proven valid for aramid fibers, while it is not feasible for fibers such as Spectra [27]. Inputting both the relevant projectile and fiber characteristics into the Buckingham Pi Theorem as described by Taylor [40], the two following dimensionless parameters were introduced:

\[
\sigma = \Pi \left( \frac{V_{50}}{U^{1/3}}, \frac{A_d A_p}{m_p} \right),
\]

(1.1)

\[
U = \frac{\sigma \varepsilon}{2\rho} \sqrt{\frac{E}{\rho}},
\]

(1.2)

with \(\sigma\), \(V_{50}\), \(\rho\), \(\sqrt{\frac{E}{\rho}}\), \(A_d\), \(A_p\), and \(m_p\) representing fiber failure stress, 50% failure velocity, fiber density, fiber elastic wave speed, fabric areal density, projectile presented area, and projectile mass, respectively. These results indicate that a single curve can be used to relate the ballistic limit normalized by the wave speed and specific energy of failure of a certain fabric or composite system to the product of the areal density of the ply and the projectile attack area normalized by the projectile mass. As can be seen by Figure 1.8, this analysis technique allows for a unique comparison of the most important impact parameters for a wide host of fiber types.
Figure 1.8. Comparison of two relevant dimensionless parameters developed by Cunniff [10].
It is important to note the limitations of this suite of dimensionless parameters and how it can over-predict or under-predict the ballistic impact limit. First, two assumptions regarding the transverse impact of the fiber are made; lateral forces do not contribute to the energy absorption and they also do not contribute to the failure of the system. The former, which is true for most fiber impact situations, is made because lateral compressive stresses are quite localized, and thus are deemed negligible. The later, which in some cases is most definitely false, removes the potential of a failure mechanism arising from this localized compressive stress, and assumes that the fiber fails only due to tension and fiber elongation. It is noted that fibers such as UHWMPE and glass will fail at a $V_{50}$ below the predicted value because of failure due to melting in the deformation zone and failure due to cracking from the lateral penetration, respectively, while Kevlar will follow the trend because it is composed of fibrils which inherently lose a minimal amount of the tensile strength when deformed in the lateral direction. The other main limitation of this parameter lies in the input mechanical properties; it does not take any high-rate effects into account. Some fibers, such as A265, have been shown to exhibit a change in mechanical properties due to a strain-hardening effect [36], and this parameter thus under-predicts the capabilities of these fibers in a ballistic environment. Regardless of these limitations, this parameter does provide an effective screening tool to better predict a fiber’s capabilities in a ballistic resistant fabric and allows designers to better understand the relevant mechanical properties that are needed to fashion an apposite ballistic fiber.

1.6 Fabric Impact Failure

The underlying reasoning high-performance fibers perform so effectively in ballistic environments stems from their ability to disperse strain energy rapidly away from the impact sight, thereby attenuating localized strains that develop upon projectile-fabric contact. Verification of this rapid strain wave velocity effectiveness has been shown both numerically [6] and from experiments [41] via altering the sound speeds in fabric
systems. Additionally, although there are numerous cross-over locations throughout the fabric system, it has been demonstrated that the longitudinal and transverse waves generated in the principal yarns are responsible for the majority of the energy dissipation away from the impact site [42] in the form of both kinetic and strain energy. Lee et al., correlated the number of broken fibers to the amount of energy which was absorbed by the fabric system, thereby concluding that principal yarn straining is the primary mechanism of energy absorption [43].

With regards to the numerous cross-over locations along the principal yarns lengths, Freeston and Claus observed that strain wave transmission and reflection at yarn-crossovers did not effect the propagation of the longitudinal wave away from the impact sight [44] That said, the cross-over junctions have been shown to play a secondary role in the fabric system. Primarily, Roylance showed that yarn cross-over affects the stress-wave profile along the principal yarns [42]. Additionally, the deflection of the principal yarns causes there to be a similar transverse propagation effects to occur in the cross-over yarns that propagate outwards in a cascade or fractal effect until the principal yarns break due to the accumulated strain being greater than the failure strain of the yarn [7]. The effect of the secondary yarns is also exemplified by the weave itself; unbalanced weaves and loosely woven fabrics have been shown to perform poorly [7] as there exists an optimal weave density [45]. Such a weave density is required so as to circumvent the ability of the projectile to ‘nose’ through the fabric panel [8]. This is further corroborated by the presence of matrix (epoxy/resin) material, which restricts yarn movement, thereby engaging more yarns and ultimately increasing the absorbed energy [46] [43]. It must also be noted that during low velocity impact, creasing and stretching is able to propagate outwards towards the edge of the panel via the transverse wave, thereby engaging the entirety of the fabric system [15,47]. Although periphery, the secondary yarns do indeed play a role in halting the incoming projectile when impacting below the ballistic limit.

Interestingly, above the ballistic limit, there is a stark change in the fabric behavior, as failure becomes both very rapid and quite localized, thereby allowing
the projectile to penetrate the panel during the initial stress-rise in the principal yarns [7, 47]. Thus, the principal stress waves do not have time to move sufficiently outwards and engage the surrounding material, or even to bounce back from the surrounding boundaries. Such an ‘immediate’ rupture has been seen by quite often by authors throughout the literature. Naming just a few, studies by Lyons [48], Cunniff [7, 12, 49], Prosser et al. [9, 11], Iremonger [13], Shim et al. [47], and Tan et al. [15], have all pointed towards an extremely localized failure wherein very little energy is pulled away from the projectile before perforation. In other words, at high enough impact velocities, the fibers/yarns will fail nearly instantaneously upon impact [48].

This localized failure is not just limited to full fabric systems; it can be seen in transverse impact of single yarns as well. At high impact velocities, various authors have found that single yarns impacted in the transverse direction exhibit extremely localized deformation and failure [50–54] and has been aptly called the critical velocity even by early authors who lacked current state-of-the-art imaging equipment [55–61].

It must also be noted that for multiple ply systems, inelastic failure can also occur below the ballistic limit. At sufficient impact velocities, local failure is promoted in the first few layers due to the impacting velocity being above the $V_{50}$ for a single ply system. Remaining layers of the system are then seen to exhibit the membrane behavior [62] or delamination and fiber pull-out. Interestingly, this local inelastic failure of the front plies dissipates such little energy that the front layers in a fabric system can be replaced with a less expensive sacrificial material, with the remaining backing fabric being composed of high-performance filaments [49]. As a final example or proof of theory, the presence of a localized failure has also been strikingly demonstrated by Cunniff who impacted fabric systems with a variety of aperture window sizes and found that the ballistic limit was influenced by the aperture size due to constraints on transverse deflection imposed by the smaller opening window. Yet above the ballistic limit, the aperture size became a negligible factor as perforation occurred with minimal transverse deflection [7].
Ultimately, the impact into a fabric system is well described by Shahkarami et al. to fit in three regimes [63]:

1. Zone 1: Low velocity impact results in the projectile completely arrested by fabric.

2. Zone 2: Medium velocity impact results in global deformation and strain energy build-up before deformation and failure.

3. Zone 3: High velocity impact results in the projectile losing only a small amount of incident velocity and failure of the fabric is highly localized.

The question then for the current work then is entirely based around understanding the transition from Zone 2 to Zone 3. Below some velocity regime, fabric is able to dissipate huge amounts of impacting energy, thus halting the projectile. Yet above this velocity, the stark change in material behavior is rooted in a change in physical deformation, namely, the fabric experiences a local failure brought on by failure of the constituent yarns and filaments. It is the purpose of this work to try and understand this local failure and to determine if the failure criterion is used throughout literature (filaments fail in pure tension) is indeed valid. Although this work is specifically geared at understanding the local failure in the fabric system, it must be remembered that true understanding and prediction of fabric performance must include both fabric geometry and material properties, which are inherently coupled together to form a structural response [6]; as ample efforts have been directed at the former, this work is ideally concerned with the latter.

It is again noted that further introductory remarks are left to each respective ensuing chapter, as it seems better suited to introduce each chapter with its respective goal, due to the compilation nature of this document.
2. Dynamic failure of Dyneema® SK76 single fibers under biaxial shear/tension

Adapted from:

2.1 Abstract

Dyneema® SK76 single fibers have been subjected to various known levels of both torsional shear strain and axial tensile stress in efforts to determine the resulting combined loading effects. High rate axial tension experiments were performed on pre-twisted fibers utilizing a miniature tension Kolsky bar and MTS servo-hydraulic system, while the resulting torque generated by fibers loaded to specific degrees of shear strain was determined via implementation of a video-based torque sensing technique. Compilation of the two stress state environments has generated a biaxial failure surface criterion yielding the residual tensile strength of single fibers when subjected to a specific grade of shear stress. Further analysis of fiber surface damage reveals longitudinal surface striation development at elevated levels of shear strain. Propagation of these undulations has been also investigated in order to determine their penetration depth and resulting internal fiber damage.

2.2 Introduction

High performance fibers are used in a vast number of applications due to their prolific material properties in axial tension. Thus, the overwhelming amount of mechanical testing of both single filaments and yarns is performed in an according envi-
ronment. However, it has been shown that these fibers possess a degree of anisotropy, and loading in multi-axial stress states, while greatly limited by fiber geometry, results in drastically attenuated material response [31,64]. Although the initial axial compression and tension moduli of single fibers have been shown to be similar [65–67], the ultimate failure stresses in both of these loading arrangements are highly contrary due to their disparate stress strain histories [67,68]. This is demonstrated by the compressive strength of polymeric fiber composites; indeed, the compression strength of the conglomerate is limited by the compressive rupture stress of the fibers themselves [69,70]. The origin of this tension compression disparity lies in the architecture of the highly drawn polymeric single filaments. It is generally accepted that these fibers possess a fibrillated microstructure, which when loaded in a compression environment, fail via chain slippage and/or fibril buckling, both leading to ultimate fibril separation [31,71–73]. The allowance of this transverse separation is a function of the extremely poor interchain interaction, consisting of solely hydrogen bonding and van der Waals attraction [74]. The resulting compressive filament failure is described by kink band formation and occurs at critical compressive stresses which are roughly an order of magnitude less than the required axial tension rupture stress for a number of polymer and pitch-based carbon fibers [75].

It is also important to note that the allowance of fibril separation requires intermolecular shear, and various authors have predicted the ultimate axial compressive strength of fibrillated polymeric fibers to be equal to the shear modulus present between the fibrils [71,76]. Using a torsion pendulum apparatus, several studies have been performed in efforts to divulge this shear modulus for a number of different fiber types [31,32,64,76–80]. Interestingly, three of the aforementioned reports [76,78,81] found that the apparent shear modulus exhibited by various high performance fibers was subject to the level of axial tension being placed on the specimen due to the torsional pendulum mass inherent in the experimental setup. A linear relationship was found to exist between the apparent shear modulus and axial load, being at-
tributed to either resistance to volume change [81] or effects rendered due to off-axis loading [76].

Furthermore, it has also been shown that the level of torsional shear strain incurred by an aramid fiber has a definite effect on the residual axial strength [31]. Kevlar 49 was able to incur a torsional shear strain up to 10% before its axial tension failure stress was noticeably reduced. Upon further increasing levels of torsional strain, a linear reduction in residual tensile strength was determined. It is also important to note that the maximum fully recoverable torsional shear strain was found to be 10% and upon reaching this level of shear, longitudinal cracks were seen to develop on the surface of said fibers. Whether the reduction in axial strength was due to propagation of surface cracks or because of multi-axial loading is unknown, but the presence of such defects was attributed to permanent molecular shear slippage. In light of this, it is necessary to observe that high-performance fibers possess a long range order, and in the case of Kevlar 49 fibers, may consist of axially oriented p-phenylene terephthalamides (PPTA) chains being transversely connected via hydrogen bonding to form radially oriented sheets, which are then held together by van der Waals attraction [74,82]. Thus it is quite possible that this long range superstructure, which predates inherent differences in fiber mechanical properties, may sufficiently alter the failure criterion under both simple and complex stress states. It has even been suggested that filament rupture when loaded in axial tension may be governed by its transverse properties [83], rendering the current understanding of fiber failure severely restricted.

Therefore, it is of upmost importance when studying high-performance fiber material properties to discern the effect of shear on the tensile failure of said fibers if they are to be used as constituents in a structure which may place them in a complex stress system. Accordingly, it is the purpose of this study to address the failure criterion of a typical high performance fiber being loaded in a biaxial tension/shear stress state. Varying levels of axial tension and torsional shear are simultaneously imposed
on UHMWPE fibers within two different loading schemes in efforts to determine a representative biaxial failure surface.

2.3 Experimental

2.3.1 Materials

Dyneema® SK76 single filaments were carefully extracted from the yarn and then mounted onto a large cardboard substrate. Ends of each filament were then coated with a thin layer of AuPd using a Hummer 6.2 sputter coater and imaged with a high resolution scanning electron microscope (HRSEM) at a 5 to 6 kV accelerating voltage in order to determine isolated specimen diameters. Single fibers were then each individually loaded onto a twisting apparatus shown in Fig. 2.1 and rotated successively to levels 7%, 14%, 21%, 28%, 35%, 42%, and 49% torsional strain in order to produce representative specimens from every single fiber sample thereby limiting inter-filament differences. These strain levels may also be described in terms of fiber helical angles, possessing values of 4.0°, 8.0°, 11.9°, 15.6°, 19.3°, 22.8°, and 26.1°, respectively. It is important to note that fibers were attached at one end to linear bearing slides in order to ensure negligible axial pre-stress due to the shortening event inherent in the twisting procedure. Following successful achievement of needed torsional strain levels, twisted specimens were then mounted onto pre-made cardboard substrates possessing cutouts of required testing gauge lengths, 5 mm and 20 mm for the dynamic tension and shear stress experiments, respectively. Fibers were then adhered to the substrates utilizing an appropriate epoxy rendering the effect of pull-out during future testing undetectable. Substrates used in the high rate tension experiment were further modified by the application of set screws running concentric with the fiber specimen, which can be seen in the inset of Fig. 2.2, while specimens intended for the torsion pendulum apparatus had a set screw attached to only one end of the substrate, as seen in the inset of Fig. 2.3. It is important to note that analysis of single filaments possessing lengths equivalent to initially untwisted fibers removed
from the yarn and then loaded into the twisting setup, possessed diameter variations along the length of at most 0.5 μm, rendering errors in strength calculations less than 8% of the correct strength values. Furthermore, at least five iterations were performed for each testing condition, aiding in the reduction of unwanted experimental error.
Figure 2.3. Experimental apparatus used for video-based torque sensing experiments
2.3.2 Uniaxial Tension

In order to accurately assess the residual tensile strength of these extremely thin fibers (∼16 μm) at both quasi-static and high strain rates, a servo-hydraulic MTS device and a miniature tension Kolsky bar, seen in Fig. 2.2, were implemented allowing for strain rates of 0.01 s⁻¹ and 600 s⁻¹, respectively. In contrast to the typical Kolsky tension bar, this miniature apparatus lacks the employment of a transmission bar, as the force from the fiber specimen is so small that the transmission bar is scarcely moving, and the bar-surface strain cannot be measured with a reasonable signal-to-noise ratio. Therefore, as described in a previous study [84], a high resolution quartz crystal load cell, which also does not move, replaces the transmission bar and the force history seen in the load cell is collected.

In efforts to monitor the achievement of a constant strain rate during the deformation cycle, a direct measurement of bar end displacement has been adopted for this study and can be seen in Fig. 2.2. A 5 mW laser diode source produces a laser fan sheet which is gathered onto a photo diode sensor by a double optical lens collection sequence. Deflection history of the incident bar end is determined by its level of protrusion into the laser line field, thereby increasing the magnitude of radiation intensity sensed by the detector during the test sequence due to the bar’s exit from the laser sheet. Both load cell and laser detector output voltage signals are synchronized and collected with an oscilloscope for future determination of force and displacement histories, which can be seen in Fig. 2.4. The incident pulse, following an initial 80 μs ramp, reaches a constant plateau lasting 100 μs, pointing towards a region of constant strain rate in the latter interval. The displacement signal shows a region of linear increase of 80 μs following ramp initiation, which too confirms that the sample has deformed and failed in a region of constant strain rate.
Figure 2.4. Typical Kolsky bar voltage signals
Figure 2.5. Strain rate history experienced by single fiber during tensile loading
From the laser detector signal, the accumulated strain ($\epsilon$) and strain rate ($\dot{\epsilon}$) histories experienced by the fiber specimen can be determined from the following equations:

$$\epsilon = \frac{d}{l_s} \quad (2.1)$$

$$\dot{\epsilon} = \frac{v}{l_s} \quad (2.2)$$

where $d$ is the bar displacement measured by the laser detector, $v$ is the bar end velocity taken as the derivative of the displacement-time history, and $l_s$ is the original sample gauge length. The aforementioned assessment of constant strain rate deformation is further corroborated by Fig. 2.5, which shows a detailed sample strain rate history in conjunction with the experienced engineering stress. A region of constant strain rate, being 600 s$^{-1}$ in value, is apparent 60 $\mu$s following the onset of bar displacement and lasts until ultimate failure of the specimen. Regarding the depicted stress history, there is no sign of fiber slippage at the glue joints, indicating that a satisfactory epoxy has been used.

Failure surfaces of twisted fibers pulled in tension have also been coated with a thin layer of AuPd and imaged with HRSEM as previously described. Fibers twisted to 21%, 35%, 42% and 49% shear strain were embedded in EPON and then polymerized at 60°C for 48 hours. Thin slices less than 100 nm were then cut with an ultramicrotome in an orientation so as to produce cross sections perpendicular to the fiber axial direction. Samples were then imaged with an FEI/Phillips CM-10 transmission electron microscope (TEM) at an accelerating voltage of 80 kV.

### 2.3.3 Shear Stress Determination

In order to determine the level of shear stress experienced by the fiber at elevated degrees of shear strain, a testing procedure has been developed which in effect acts as a fiber torque sensor and can be seen in Fig. 2.3. As previously described, fiber spec-
imens have been twisted to 14%, 21%, 28%, 35%, 42%, and 49% and then laid onto cardboard substrates. The described fiber substrates are then threaded on one end into a thin torsion disc. The substrate-disc assembly is loaded into the experimental apparatus by mounting the other end of the substrate to the rigid fixture with an appropriate clamping mechanism and then both sides of the cardboard are carefully cut ensuring minimal jostling of the fiber itself. The pendulum disc is then allowed to freely rotate and a plastic housing is placed over the entire apparatus in order to reduce the effect of ambient air flow. Videos of the rotation history are recorded and later analyzed in order to determine the angular acceleration history exhibited by the disc, thereby rendering access to the level of shear stress present in the fiber, which is described below. Seven different sized torsion discs have been used in order to determine the effect of axial stress on the apparent shear stress present in the fiber at the varying levels of shear strain. Discs were selected so as to load the tested fibers up to levels ranging between 10%-80% of their tensile breaking force. It is also important to note that in order to ensure a minimal amount of test damping, torsion discs possessing a large moment of inertia have been used [78,80], and the disc itself has been machined from 6061 aluminum, in efforts to negate magnetic effects. Due to the negative effects of varying environmental conditions and fiber creep [32,77,85], tests were performed at a temperature and relative humidity of 22.5 °C and 34%, respectively, ensuring that the delay between twisting and testing was no more than 24 hours, being necessary for appropriate epoxy curing time. It is also important to note that the 20 mm gage length was implemented in the shear stress test for ease in the experimental procedure (e.g. substrate cutting, alignment), but as it is too long for the dynamic tension test, 5 mm gage lengths were used for the Kolsky bar experiments in order to promote the achievement of dynamic equilibrium.
2.4 Results and Discussion

2.4.1 Uniaxial Tension

Effects of gage length on untwisted fiber strength has been performed at various strain rates. As can be seen in Fig. 2.6, gage lengths of 5 mm, 10 mm, and 100 mm were pulled in tension at $10^{-3}$ s$^{-1}$ and $10^{-2}$ s$^{-1}$ using the described MTS servo-hydraulic testing system, ensuring an acquisition of at least 15 valid test results. For all tested gage lengths, there may exist a direct positive correlation between tensile strength and testing strain rate, but results are not conclusive, as this sensitivity lies within the statistical variation inherent in the single fiber mechanical response. Gage lengths of 5 mm and 10 mm were then pulled in tension at both 400 s$^{-1}$ and 600 s$^{-1}$ on the aforementioned Kolsky bar apparatus. While the filament tensile strength at these two higher strain rates increases and decreases with increasing strain rate for the 10 mm and 5 mm specimens, respectively, there appears to be a slight increase in tensile strength from the quasi-static to high rate regime. This positive correlation between strain rate and tensile strength has been evidenced in previous works [27, 86, 87] examining UHMWPE fibers, albeit the previous results depict a more acute strain rate sensitivity. It can also be seen in Fig. 2.6 that both 5 mm and 10 mm gage lengths exhibit very similar strength results to each other at all four tested strain rates and it is apparent that both short gage lengths resulted in reduced failure stress levels than the longer 100 mm gage length during quasi-static tests, affirming the lack of length effects for the tested gage lengths which commonly govern single fiber tensile response [87–89]. It has been conjectured [87] that this lack of gage length sensitivity can be explained by the flaw distribution present along the single fiber length, which in the case of this testing environment, suggests that critical defects have a period spacing less than 5 mm. It is also important to note that 100 mm samples were not pulled in dynamic tension, as such long gage lengths would be unable to reach dynamic equilibrium during the elevated loading rate regime.
Following gage length and rate sensitivity analysis on zero shear strain filaments, single UHMWPE fibers were twisted to varying levels of shear strain and then loaded in both quasi-static ($10^{-3}$ s$^{-1}$) and high-rate (600 s$^{-1}$) axial tension with a servo-hydraulic MTS and the Kolsky bar setup shown in Fig. 2.2. Resulting failure stress levels as a function of maximum torsional strain experienced on the fiber surface are represented in Fig. 2.7 for both low and high strain rates in red and blue, respectively. The tested single filaments retain their native axial tensile strength up to at least a torsional strain level of 14% where upon further twisting there is a clear decrease in the residual tensile rupture stress with the reduction looking linear in nature for both the low- and high-rate conditions. Interestingly, the tensile strengths of fibers pulled
at a high-rate are less effected by increasing torsional strains than their quasi-static counterparts, thereby suggesting the presence of a rate sensitivity in twisted single fibers being much more demonstrative in nature than the minimal strain rate effect found in the untwisted results previously described.

The linear quasi-static tensile strength degradation as a function of increasing torsional strain has previously been found to exist for Kevlar 49 fibers, where the loss in residual quasi-static tensile strength was deemed negligible up to torsional strains of 10% [31]. Further twisting caused the residual strength to fall off in a linear fashion, with the degradation being attributed to either multi-axial loading or cracking on the fiber surface due to intermolecular shear slippage resulting from the inherent poor transverse bonding. In the current work, surface striations were also uncovered on the surfaces of UHMWPE fibers twisted to high levels of shear strain, which can be seen in Fig. 2.8. Both un-tensioned fibers and fibers pulled with the aforementioned Kolsky bar apparatus show similar striation evolution being described by increasing undulation prevalence and accentuation with increasing shear strain. It may be postulated that if these longitudinal undulations do represent surface cracking, the UHMWPE fiber possibly possesses a higher level of circumferential bonding strength than the inter-sheet Van der Waals attraction exhibited by the Kevlar 49 fiber [82], as the Dyneema® fibers do not exhibit surface crack damage until a shear strain between 14-21%, being opposed to the Kevlar 49 fiber which has been found to exhibit longitudinal splitting at strains between 8-15% [31].

In order to discern if these surface bands penetrate into the core of twisted fibers, TEM work has been performed on fibers cross sections twisted to 21%, 35%, 42%, and 49% shear strains. The resulting images shown in Fig. 2.9 depict no evidence of the aforementioned surface cracks penetrating deep into the fiber core. The undulating striations represent artifacts inherent in the TEM sample preparation technique wherein folding of the thinly sliced cross sections occurs due to ineffective infiltration and polymerization of the fiber UHMWPE fiber cross section. It is important to note that the fibers used in the TEM analysis were twisted to the prescribed level of shear
strain, while ensuring a minimal level of applied axial stress. Therefore, while the introduction of a tensile axial load may encourage surface crack penetration, it is clear that the application of a shear strain itself does not promote the longitudinal surface striations to penetrate into the fiber inner core and the residual tensile strength of twisted fibers may be effected not by inherent twisting damage, but rather possibly by multi-axial loading.

2.4.2 Shear Stress Determination

In order to determine the level of shear stress experienced by single filaments when loaded to varying degrees of torsional shear and axial tension, a video-based fiber torque sensing device has been employed. The rotational behavior of the torsion disc apparatus can be described by the following:

\[ \sum T = I_{\alpha} \quad (2.3) \]
Figure 2.8. Surface topography for both un-tensioned (a-f) and post-tensioned (h-l) single fibers twisted to appropriate levels of shear strain. Surface striation evolution appears similar for both environments being described by increasing undulation prevalence with advancing shear strain.
Figure 2.9. TEM cross sectional images show no penetration of afore-
mentioned surface cracks into fiber core. Undulating striations are
artifacts inherent from the sample preparation which did not include
the polymerization phase typically used in the biological ultramicro-
tome preparation.
where $T$ represents the various torques present in the torsion disc, $I$ signifies the disc moment of inertia, and $\alpha$ represents the disc angular acceleration. As previously mentioned, torsion discs possessing large moments of inertia were used in order to ensure long periods of oscillation, thus rendering the effect of damping negligible [78,80]. Therefore, it can be assumed the only appreciable torque acting on the disc is derived from the twisted fiber from which it is hung. The torque exhibited by the fiber on the disc is thus related to both the disc geometry and angular rotation, $\alpha$, by:

$$T_{fiber} = \frac{mD^2}{8} \alpha \quad (2.4)$$

where $m$ is the disc mass and $D$ is the disc diameter. The total torque exhibited by the fiber may also be described as a summation of all infinitesimal moments, $dM$ produced over the fiber cross sectional area, $A$, as:

$$T_{fiber} = \int_A dM \quad (2.5)$$

Assuming that the apparent shear stress distribution is a linear function dependent on radial location, Eq. 2.5 may be described by:

$$T_{fiber} = \int_0^{2\pi} \int_0^r \tau_{max} \frac{\rho^3}{r} d\rho d\theta \quad (2.6)$$

where $\tau_{max}$ represents the shear stress present on the fiber surface located at the fiber radius $r$. When integrated over the fiber cross sectional surface using polar coordinates signified by $\rho$ and $\theta$ and then simplified to include fiber diameter $d$, the total torque produced by the fiber can be described as

$$T_{fiber} = \frac{\pi}{16} d^3 \tau_{max} \quad (2.7)$$

Equating Eqs. 2.4 and 2.7, the following relationship can be found for $\tau_{max}$:

$$\tau_{max} = \frac{2mD^2}{\pi d^3} \alpha \quad (2.8)$$
Thus, the only term needed to be resolved from the rotation history of the fiber experiment is the angular acceleration, $\alpha$, which is determined as the slope of the angular velocity-time curve found from the recorded video sequences.

In light of this novel torque sensing device, fibers were subjected to known levels of both torsional shear strain and axial tensile stress and by assessing the rotational accelerations experienced by the fibers under the varying loading conditions, resulting shear stresses were determined with Eq. 2.8, which can be seen in Fig. 2.10. At specific levels of shear strain, the apparent torsional shear stress increases in a linear fashion with increasing applied axial tension, therefore linear fitting functions have thus been overlaid. From Fig. 2.10 it can also be seen that with increasing levels of shear strain, there is a likewise increase in apparent shear stress at constant axial tensions. This is further demonstrated by the increasing slope of the fitted linear functions with increasing shear strain levels, resulting in apparent shear stresses being increasingly more sensitive to applied axial tensions at elevated shear strains.

It is also interesting to note that the loss in recoverable shear strain was negligible, even at the highest shear strain value of 49%, wherein the magnitudes of untwist and re-twist were equivalent during the first oscillation period. This is in contrast to previous work where appreciable levels of plastic torsional deformation was exhibited by Nylon and Kevlar 49 single fibers twisted to high levels of torsional strain [31, 90]. Only at small strains, roughly 10% for Kevlar 49 filaments, was there solely elastic deformation. Also, the fully elastic deformation of the UHMWPE fiber further corroborates the previously made assumption of negligible test damping.

### 2.4.3 Biaxial Failure Surface

Due to the linear shear stress vs. axial tensile stress response exhibited by fibers twisted to a specific level of shear strain, it is reasonable to extrapolate this curve to the tensile failure stress levels found from the high rate tension tests represented in Fig. 2.7 for fibers loaded to the appropriate shear strains. This is represented by the
starred markers in Fig. 2.10. In order to provide increased clarity of the effect of the apparent shear stress on the failure tensile strength, these extrapolated values are then accordingly positioned in a tensile failure stress vs. apparent shear stress plot, shown in Fig. 2.11. From the generated curve, it appears that the axial tensile stress of this UHMWPE fiber is unaffected by the apparent shear stress up to levels of 1 GPa. Further application of shear stress causes a substantial reduction in the residual tensile strength, indicating that high performance fiber tensile strength is a function of shear stress, which is currently neglected in fiber composite modeling [7,8,91]. Therefore, to the authors’ knowledge, the culmination of experimental work from Figs. 2.7 and 2.10 into Fig. 2.11 represents the first exploration into the effect of combined biaxial tension/shear on the failure of single fiber filaments. Furthermore, it may provide an introductory testing environment in which the feasibility of introducing possible polymeric fibers into ballistic panels can be determined for applications such as body armor and turbine fragment containment.
Figure 2.11. Biaxial failure surface criterion of single Dyneema® filament when loaded in torsion-tension environment.
2.5 Conclusion

Single Dyneema® SK76 fibers have been simultaneously subjected to varying levels of both axial and shear stresses, revealing a clear affiliation between the coupled stress states. Indeed, a definite biaxial failure surface has been generated showing the resulting tensile failure stress of these highly anisotropic fibers when loaded to a specific level of shear stress. Residual tensile strengths have been shown to be unaffected by the apparent shear stress and shear strain up to 1 GPa and 14%, respectively. At increased levels of shear strain, resulting fiber surface damage has been analyzed with HRSEM, revealing the appearance of longitudinal striations, which are attributed to the low degree of transverse bonding, being replete in the majority of highly oriented polymeric fibers. Internal core damage of the twisted fibers has been further investigated with TEM analysis of fiber cross sections, depicting negligible propagation of the surface undulations into the fiber core. Finally, it has been suggested that the evolution of the aforementioned biaxial failure criterion may lead to a possible initial testing regime in which the efficacy of implementing a potential fiber candidate in a ballistic environment could be determined before fabrication of the entire composite system. Furthermore, the clear effect of the applied shear stress on the resulting axial tensile strength of this high performance fiber may introduce an additional component into current fiber system modeling.
3. Degradation of yarns recovered from soft-armor targets subjected to multiple ballistic impacts

Adapted from:

3.1 Abstract

Post ballistic impact residual yarn mechanical properties were analyzed from two different as-received shoot packs composed solely of AuTx yarn possessing the 2/2 twill weave structure, one being impacted by 9-mm projectiles and the other by 2-grain projectiles. It was found that yarn mechanical properties from both shoot packs yielded similar results, regardless of yarn orientation, ply location, or penetrator size, which indicates that ballistic damage in the packets is very localized, producing little damage to the neighboring yarns. Mechanical properties of these woven, ballistically impacted, and then extracted yarns were compared to as-received native spooled AuTx yarn yielding a slight reduction in tensile strength, an increase in failure strain, and a reduction in elastic modulus, thereby yielding little variation in yarn toughness.

3.2 Introduction

Fabric panels woven from high performance fibers and yarns are increasingly employed in high strain-rate applications such as soft ballistic armor, turbine fragment containment systems, and anti-spall linings. It is well known that many parameters play a role in a soft armor system’s projectile halting efficiency including but
not limited to fiber-fiber friction, yarn-yarn friction, fiber-projectile friction, fabric weave structure, fiber/yarn mechanical properties, projectile impact velocity, projectile impact obliquity, projectile nose geometry, and projectile material properties [7,8]. Indeed, vest penetration is an extremely complex process, but as an initial screening procedure, yarn mechanical tensile properties can be used to indicate possible soft-armor performance with a high degree of success [10,92]. Therefore, it is of great use to analyze yarns having undergone various treatments or aging processes in order to better predict overarching fabric performance having been exposed to field-use by armed personnel.

Previous works have aimed at understanding yarn tensile behavior, both from an experimental approach and/or statistical modeling [93–103], thereby shedding much light on both yarn testing procedures and resulting deflection-load response. Although single fiber mechanical property characterization is of great importance in understanding the possible effectiveness of a given fiber type in ballistic applications, once a filament material is deemed feasible for use, yarn rupture analysis becomes of extreme interest when trying to understand energy dissipative capabilities of an armor system due to yarn seizing and scouring procedures. Furthermore, even though previous aramid single fiber testing has revealed little rate dependence and possible sample length effects on resulting mechanical properties [35,36,89,97,104], there exists a much more demonstrative variation in mechanical response on yarns via said effects due to the structural failure phenomenon of fiber breakage and load reorientation [93–96,100], thereby yielding a definite increase in stress due to an increase in testing strain rate or a decrease in the yarn gauge length [98,99,101–103]. Thus, in order to correctly uncover the key issues affecting yarn tensile behavior, it is imperative to accurately determine appropriate testing parameters which will not mask the effect of the degenerative issue being studied.

It is well known that fabric processing procedures such as yarn seizing, weaving, and scouring can alter the effectiveness of an armor system, and these effects should be fully understood before vest production [102]. Furthermore, environmental processes
such as UV-degradation, humidity, temperature, or abrasion, which are all reasonably
difficult to properly reproduce in lab environments, can each drastically alter the
ultimate halting capabilities of an armor system [22, 28, 105, 106]. While each of these
issues has been analyzed in some detail, the effect of multiple loadings on a fabric
system has been much less established. In many circumstances, these ballistic fabric
panels may undergo several impact traumas, requiring the fabric to retain nearly all
of its threat arresting capabilities during repetitive invasive strikes. Thus, while it is
imperative to pursue the development of robust armor systems which are capable of
withstanding long-term environmental aging, it is also of importance to ascertain an
understanding of damage mechanisms of said systems due to stress wave propagations
throughout the surrounding fabric structure during repeat loading activity. In light
of this current lack of understanding in vest behavior, it is clearly of use to analyze
single yarns removed from shoot packs having undergone repeat invasion in order to
understand the degradative effects of impact on surrounding fabric zones. Therefore,
two shoot packs composed of AuTx ballistic fibers, which were previously impacted
by multiple projectile threats, have been investigated in order to determine residual
mechanical properties of yarns being located within various proximity of impacting
points. First and foremost it is the goal of this study to establish the viability of the
AuTx system, but in addition it is hoped that results of this work may also shed light
on the consequence of multiple impact loadings in typical soft armor from varying
sized projectiles.

3.3 Materials and Methods

Two similar AuTx shoot packs have been received for post impact analysis, each
being composed of AuTx high-performance aramid fibers. Received shoot packs,
SPA and SPB, were previously impacted with 2-grain flat head and 9-mm ball round
projectiles, respectively. Similar shoot patterns were used on both SPA and SPB
and can be found in Figure 3.2. Each shoot pack consisted of 34 soft fabric plies,
being woven in the 2/2 twill format depicted in Figure 3.3. The warp-weft directions of the ply weave were found to randomly lie in either the 0 or 90 degree alignment with respect to the vertical direction of Figure 3.3 and 3.3. The orientations of the included plies from both packs can be seen in Table 3.1.

Determination of the warp and weft yarn directions were decided by optical analysis of the color gradient found between adjacent yarns, as the warp yarns are typically pulled from different yarn bobbins while the weft yarns consist of a continuous strand.
Figure 3.2. Shot patterns used for SPB.
Figure 3.3. 2/2 twill weave structure found in received AuTx shoot packs.

Table 3.1. Orientation of plies within the shoot pack.

<table>
<thead>
<tr>
<th>ply level</th>
<th>2-grain shoot pack (SPA)</th>
<th>9-mm shoot pack (SPB)</th>
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<tr>
<td></td>
<td>Ply level</td>
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which is oscillated back and forth perpendicular to the warp direction by an undulating trundle. Therefore, the warp direction is expected to possess a greater color variation between adjacent yarns than their weft counterparts. It is important to note that the usual method of establishing warp and weft directions by performing yarn pull-out tests [107], by measuring crimp amplitude [102], or by resolving differences in yarn diameters [108] yielded inconclusive findings on this material, thus the aforementioned color gradient technique was utilized. Verification of this procedure is bolstered by the slight variation in warp-weft stress-strain history, which is described in Section 3.4. It is important to note that the the last ply (ply 34) of SPA was initially sacrificed to create both pull-out specimens and yarn crimp samples, thus yarn tension tests from this panel could not be performed.

From each ply within the shoot packs, multiple single yarns were pulled in a strategic pattern from both the warp and weft directions, and the pull pattern can be seen in Figure 3.4. Three yarns were extracted between adjacent rows and columns of impact points in an effort to determine the effect of yarn-shot proximity. Spacing between both adjacently pulled yarns and yarn-impact points was 9-mm, but it is important to note that some shot placements were not ideal, resulting in variation of the actual yarn-shot spacing. Extracted yarn specimens were then mounted onto large foam boards for storage until future testing, ensuring that minimal amount of twist was incurred in efforts to minimize the apparent strengthening effect of low level yarn twist [22,109]. It is thought that low levels of twist increase filament transverse pressure, allowing broken filaments to more effectively transmit longitudinal loading via skin-skin friction [110]. Non-woven untwisted yarns which were previously received for single fiber testing were also pulled in tension in order to determine the baseline yarn mechanical properties.

Yarn cross-sectional area has been calculated as a compilation of the cross-sectional areas of its filament components. In order to ascertain the number of filaments present in a yarn, single yarns were embedded in epoxy and then cut and polished so as to count the number of single fibers present in the cross section. It was found that a
Figure 3.4. Yarns were pulled between the shot rows and columns in a fashion so as to determine the effect of yarn-shot proximity on both warp and weft yarns. (a) SPA and (b) SPB. Yellow circles represent locations where upon visual inspection, there was no penetration through the final ply, and red circles represent locations where there was full final ply penetration.
typical AuTx yarn consists of roughly 290 single filaments with each possessing a circular-cross section. Using a FEI Nova high resolution SEM, fiber cross-sections were measured to possess an average diameter of 10.22 μm. Therefore, a typical yarn cross-sectional area \( A_{yarn} \) is defined as the total cross-sectional area of all present fibers, resulting in a yarn area of 0.024 mm\(^2\). Determination of fabric denier was performed via weighing single lengths of yarn pulled directly from the spool. Post the initial mass measurement, said yarns were then placed in a 150 °C oven for three hours in order to remove any residual moisture content. Pre- and post- baked yarns exhibited deniers of 274.6 den and 266.1 den, respectively. Further heating caused insignificant losses in linear density of the tested yarns.

The previously described single yarns were then pulled in uniaxial tension via guidance of the ASTM standards (D2256) [111] and (D7269) [112]. All specimens were loaded into an MTS 810 servo-hydraulic testing apparatus outfitted with MTS Pneumatic Cord/Yarn/Wire 180° bollard grips and an Interface 1500ASK 200lb load cell, which can be seen in Figure 3.3. In order to neglect possible strain rate and gauge length effects [98, 99, 102, 103, 110, 113], specimens possessing a nominal length of 280 mm were pulled until failure at a constant strain rate of 0.01/s using the slack start procedure described in the ASTM (D7269) standard. In order to ensure negligible yarn-clamp interface slippage, clamp grip surfaces were covered with carbon tape and all specimen ends were coated in rosin powder prior to tensile testing. Only specimens breaking in the gauge section have been included. Voltage signals from the load cell and testing system were simultaneously recorded onto an oscilloscope at a sample frequency of 250 Hz after having been filtered with a 10 kHz low-pass filter. Resulting waveforms were further analyzed in order to determine failure stress \( \sigma_f \), failure strain \( \epsilon_f \), initial elastic modulus \( E_i \), and toughness. Failure stress is defined as:

\[
\sigma_f = \frac{F_{fail}}{A_{yarn}}
\] (3.1)
where $F_{\text{fail}}$ is the rupture force of the yarn and $A_{\text{yarn}}$ represents the previously described yarn cross-sectional area. Strain in the yarn is measured from crosshead displacement, therefore compliance correction for the system deflection is determined as described in ASTM standard (D7269-11) [112]. Finally, the initial elastic modulus is calculated by the tangential slope of the compliance corrected stress-strain curve immediately following yarn uncrimping, and yarn toughness is defined as the area beneath the resulting stress-strain response.

In efforts to further assess the degradative nature of the weaving process and the effect of ballistic impact on the residual strength of AuTx yarns, single fiber quasi-static tension tests have been performed on fibers removed from the as-received spool yarn. Filament gauge lengths of 5 mm, 10 mm, 50 mm, and 100 mm were carefully removed from the bundle and were pulled at 0.001/s and 0.01/s strain rates following the ASTM:1557 standard [114]. Samples were attached to cardboard sub-
strates and mounted in clamps fixed into the same servo-hydraulic MTS 810 device used for yarn tension testing. Tensile loading was directly measured via an Interface 220.24N (50lbf) load cell mounted on top of the specimen fixture and the machine output signal containing displacement measurement was used instead of traditional extensometer measurement in efforts to minimize torque effect placed on the load cell. The displacement readout derived from the machine output was calibrated with a high-resolution Epsilon extensometer. Again, both displacement and force readouts were collected via oscilloscope at a frequency of 250 Hz, with the former signal being filtered through a 10 kHz low-pass filter and the latter signal coming directly from the MTS machine output.

3.4 Results

Both SPA and SPB were initially examined at the impact locations depicted in Figure 3.4 through the entirety of the shoot pack thickness in order to determine the success rate of projectile halting after repetitive impact strikes. The figure demarcates full penetration through all 34 plies of the fabric with red markers, while projectiles which were halted within the pack have been labeled with yellow markers. Of the 17 strike points incurred by SPA via the 2-grain flat nose projectile, ten locations were fully penetrated through the entirety of the pack. No trend was found to relate the halting capability of the vest and strike order. In Figure 3.3, it can be seen that for SPB, which was impacted with the 9-mm projectile, only three of the 16 strikes were able to fully penetrate the pack. Again, no correlation was found to exist between strike location or order in relation to the halting capability of the pack. It is important to note that upon receiving both packs, the corners of each showed a folding of several of the fabric plys to an extent that some impact sites did not fully engage all ply levels. Surprisingly, some of these “fold-over” locations were still successful in halting the projectile threat.
After penetration analysis, single yarns were extracted from both SPA and SPB and then pulled in quasi-static tension as described in Section 3.3. A set of stress-strain histories from both vertically and horizontally positioned yarns within layer 1 of SPB can be found in Figure 3.6. Only this set of curves is provided as almost all samples tested exhibited a similar load history response. It can be seen from this set of stress-strain response that the weft yarns exhibit a slightly higher stress history than the warp yarns, as the latter are more detrimentally affected by the weaving process [102]. The warp yarns are pre-tensioned during the weaving process and are abraded by the oscillating trundle and trailing weft yarns, which thereby cause surface damage to the warp yarns via kink band formation or surface fibrillation.

As the complete quasi-static mechanical properties from yarns pulled from the two different shoot packs are astronomical in size, results have been reorganized for ease of visualization and can be seen in Figures 3.7,3.8,3.9,3.10. All said figures contain
results from both shoot packs in order to ascertain the effect of projectile size on residual yarn tensile properties. Figure 3.7 demarcates the residual tensile properties of the horizontal and vertical yarns from both SPA and SPB when averaged for each ply level. Figure 3.8 depicts the residual tensile properties of the horizontal and vertical yarns from both SPA and SPB when averaged through all plys within a respective pack at the pre-defined specific yarn locations. As previously stated, the orientation of the plys within both shoot packs was not consistent, and no pattern of ply orientations was determined. Therefore in order to see the effect of warp versus weft yarn, the aforementioned data was reorganized and plotted in both Figures 3.9 and 3.10. The former figure shows the effect of ply level on the residual tensile properties of yarns oriented in warp and weft directions for both shoot packs, while the latter figure depicts the effect of yarn location on residual tensile properties for both warp and weft yarns when averaged through the plys at the pre-defined specific locations. Again, these figures show both the SPA and SPB data in order to better ascertain the effect of projectile size. It is also important to note that all mentioned figures also possess the appropriate tensile property of native yarns, coming directly from the as-sent spool, yielding results unaffected by weaving, scouring, ballistic loading, or removal from the ply.

Broken fiber ends originating from yarns pulled in quasi-static tension were imaged with a FEI Nova high resolution SEM and typical fractured ends can be seen in Figure 3.11.

In efforts to further understand the effect of yarn seizing, fabric weaving, and ballistic impact, single fibers have been pulled in quasi-static tension as previously described in Section 3.3. Fibers possessing gauge lengths of 5 mm, 10 mm, 50 mm, and 100 mm were pulled at 0.01/s and 0.001/s strain rates in order to determine both length and strain rate effects on tensile response. As can be seen in Figure 3.12, there is a clear decrease in fiber tensile strength with increasing gauge length, while the effect of strain rate is inconclusive.
Figure 3.7. Effect of ply level on average horizontal and vertical yarn tensile mechanical properties within the respective ply.
Figure 3.8. Effect of yarn location on average horizontal and vertical yarn tensile mechanical properties averaged through all shoot pack plies. Yarn locations can be discerned via the numbering scheme presented in Figure 3.4. Each presented stress value is an average of strengths gathered from yarns which have been pulled from the respective displayed location from each ply through the shoot pack thickness.
Figure 3.9. Effect of ply level on average warp and weft yarn tensile mechanical properties within the respective ply.
Figure 3.10. Effect of yarn location on average warp and weft yarn tensile mechanical properties averaged through all shoot pack plies. Yarn locations can be discerned via the numbering scheme presented in Figure 3.4. Each presented stress value is an average of strengths gathered from yarns which have been pulled from the respective displayed location from each ply through the shoot pack thickness.
Figure 3.11. SEM images from fibers originating from yarns pulled and ruptured in quasi-static tension (a-b) and single fibers pulled in tension being extracted from as-received spool yarn (c-d).
Figure 3.12. Single fiber tension on AuTx fibers possessing gauge lengths of 5 mm, 10 mm, 50 mm, and 100 mm at both 0.01/s and 0.001/s strain rates.
3.5 Discussion

From the typical stress-strain histories found in Figure 3.6, it is apparent that these AuTx yarns follow the usual loading shape history typical of yarns composed of high-performance fibers [98,99,102,103,110,113]. There are four definite regions which can be described by uncrimping, linear elastic elongation, fiber breakage and inter-fiber slippage, and ultimate yarn rupture. The first region, being described by the initial J-shaped load extension, is caused by uncrimping of the yarn as it possesses an initial geometric undulation due to the weaving process and must be extended in order to pull the yarn taught [110], which generally occurs at a 0.5% level of strain. This uncrimping region is then followed by an elastic zone where the stress-strain history in the yarn is linear and lasts to a strain level of 1.5%. The tangential modulus of the stress-strain curve then gradually reduces, as various filaments within the yarn begin to break and broken fibers begin to slide past neighboring fibers, dissipating energy through friction interactions. Finally, the yarn undergoes ultimate rupture, which is defined by a sudden drop in the extension history to a negligible level of measurable load. It is important to note that even after final rupture, the yarn still looks visually intact, and the load transfer is solely defined by minimal fiber sliding interactions, being indistinguishable from a zero load output condition from the load cell. For clarification, each of these transitions zones from the presented results are described in Figure 3.6.

All tested stress-strain histories from all plies within both SPA and SPB were very similar indeed, as can be seen from Figures 3.7 to 3.10. The resulting yarn tensile properties surrounding impact zones do not look to be a function of either yarn location, ply level, or even the size of the penetrator. It is apparent that multiple repeat loading threats from widely different projectile sizes does not effect the tensile properties of yarns at different locations surrounding the impact zone. Regardless, it is important to note the 2-grain projectile was more successful in penetrating the shootpack, as full penetration was visually determined post-mortem for impact sites 1,
4, 5, 7, 9, 10, 11, 12, 14, and 17, while the 9-mm projectile was only able to penetrate three locations, being 5, 6, and 7. Full penetration locations are represented in Figure 3.4 by red circles. It is suggested that the 2-grain projectile was much more adept at “nosing” through the yarn weave, as surrounding yarn damage was minor, thereby suggesting that fewer yarns within the fabric system were engaged during the loading cycle. This is in contrast to the 9-mm projectile penetration sites, which showed much more surrounding yarn damage as impact zones were accompanied by a large amount of fiber/yarn breakage, indicating that many yarns were detrimentally involved during the impact loading. Impact sites from the first ply level of both SPA and SPB are shown in Figure 3.13. It is also important to note that 9-mm projectiles looked to be typical lead core ball nose rounds showing large amounts of plastic deformation post impact, while the 2-grain flat nose penetrators were made of steel, thereby exhibiting very little amount of post test plastic yielding. It is also important to note that upon visual tracking of impact points through the shoot pack, plies from SPB were found to exhibit much more ply sliding, even to the point that impacts 1, 9, and 10 did not look to even engage plies 1-12, as these plies looked to have slid out of the upper corners of the pack during the actual penetration testing. Interestingly, at these impact points there was no final ply penetration, resulting in successful penetration halting.

In efforts to compare the yarns which had been woven, ballistically loaded, and then extracted from the ply to native, untainted yarns, several as-received spooled yarns were pulled in tension and resulting mechanical properties can be seen overlain on Figures 3.7 to 3.10. It was found that while tensile strength is only slightly decreased by the three aforementioned processes, the failure strain increased and the elastic modulus decreased, resulting in an unaltered yarn toughness. One possible explanation of the increase in failure strain and decrease in elastic modulus is that either weaving, ballistic loading, or yarn extraction caused there to be a reduction in the yarn seizing effects or surface treatments which keep the individual fibers from easily sliding across one another. This seizing degradation can therefore allow there to be greater fiber slippage, resulting in a decreased modulus and increased failure.
Figure 3.13. Impact site for (a) SPA being hit by the 2-grain projectile and (b) SPB being hit by the 9-mm projectile. The 9-mm impact shows more yarn damage and surrounding yarn interaction, while the 2-grain impact looks to exhibit more of a “nosing” through the fabric. Both images represent impact site which showed full penetration through the entire shoot pack.
strain, as individual fibers are able to more freely slide longitudinally with respect to each other. It is also suggested that these three processes do not substantially break the individual fibers, as there is not a major loss in yarn tensile strength, which is almost entirely derived from breakage of constituent fibers.

Typical rupture morphologies of fibers originating from shoot pack yarns pulled in quasi-static tension can be seen in Figure 3.11, along with fracture morphologies of single AuTx fibers extracted from native spool yarns and pulled in tension. All imaged fibers showed fibrillar rupture morphologies, being the typical mode of fracture for high-performance fibers [37,38], and similar to the aforementioned stress-strain behavior, no stark contrast could be made for any of the imaged fibers, thereby again pointing to the similarity in minimal degradation incurred by both shoot packs.

Due to the similar nature of the yarn tensile properties exhibited by both SPA and SPB post impact, it is suggested that the repetitive ballistic impact occurrences seen by said fabrics played little role in surrounding fabric degradation, as widely different projectile sizes were used on both identical soft armor systems. Furthermore, as previously discussed and shown in Figures 3.7 to 3.10, no significant variation in yarn tensile properties was found to exist for specific yarn locations averaged through the fabric thickness and for said properties averaged over yarns within each ply level compared to neighboring ply levels. Ideally, a pristine fabric ply is needed for testing in efforts to factualize this hypothesis, but due to the nature of this testing procedure, none such fabric was provided.

Finally, in order to better assess the effect of seizing, scouring, and weaving on the strength of AuTx yarns, single filaments were pulled from the as-received spool and were tested in single fiber tension. Gauge lengths of 5mm, 10mm, 50mm, and 100mm were pulled at rates of 0.001/s and 0.01/s in order to also assess the gauge length and rate effects. As can be seen in Figure 3.12, for both 0.01/s and 0.001/s strain rates there is a definite length effect on fiber strength, as single filament strength is reduced from roughly 4 GPa to 2 GPa for 5 mm and 100 mm gauge lengths, respectively. Surprisingly, the low tensile strength of the 100 mm fibers when pulled
at both tested strain rates was less than the strength exhibited by the 280 mm yarns. It is important to note that the removal process of AuTx yarn, while performed quite delicately, was still very tedious due to the seizing present on the yarn. Thus, the actual extraction process of single fibers from the yarn must have provided additional defects to the fiber surface, thereby yielding decreased rupture strengths of the single filaments.

3.6 Conclusion

Two different as-received shoot packs, one being impacted by 2-grain projectiles and the other by 9-mm projectiles, were analyzed in order to determine residual yarn tensile mechanical properties in locations surrounding the various impact zones. Tensile strength, failure strain, elastic modulus, and yarn toughness were found to be similar for all tested yarns regardless of yarn location, ply level, or penetrator size. These results were further analyzed by comparison to native as-received spooled yarn in order to determine the effect of the weaving, ballistic impact, and extraction procedure. It was determined that the yarn strength was only slightly reduced by the aforementioned sequence of processes, while the failure strain and elastic modulus, increased and decreased, respectively, yielding a similar rupture toughness similar to the native yarn. Thus, surprisingly, it is proposed that the impact phenomenon imparts very little degradation on yarns away from the impact site. Visual inspection of the shoot packs was performed in order to determine the number of full projectile penetration sites, being ten and three for the 2-grain and 9-mm packs, respectively, which may be expected, as the 2-grain projectile is much more adept at nosing through the 2/2-twill weave structure than its 9-mm counterpart. Therefore, it can be said that the AuTx shoot packs are able to undergo repetitive striking threats with negligible damage to surrounding fabric sites and that this weave structure is much more adept at halting large caliber penetrators in comparison to small projectiles being found in shotgun rounds or fine fragmentation. In light of this behavior, it can be suggested
that the effectiveness of this armor system, and possibly most soft armor systems, at halting multiple loading strikes impacting in various locations of the armor system is due to the minimal degradative effect that an impact load has on neighboring fabric regions, most likely due to the inherent linear elastic nature of high-performance fibers. Furthermore, this positive attribute exhibited by soft-armor to withstand multiple loadings may be seen to produce positive psychological effects. Namely, law enforcement personnel or soldiers wearing such vests should know they can be shot multiple times throughout their soft armor system without the system loosing halting effectiveness against penetrators for which it is designed to stop.
4. Why the Smith theory over-predicts instant rupture velocities during fiber transverse impact

Adapted from:

4.1 Abstract

The effect of multi-axial loading on single Kevlar® KM2 fibers is explored with an emphasis on correlating the results to fiber/yarn transverse impact. A 0.30 caliber fragment simulation projectile (FSP) is slightly modified to act as a transverse loading indenter. Fiber failure angles are forced between \( \sim 0^\circ \) and \( 50^\circ \) in efforts to deduce the deleterious effect caused by such angles. Said angles are also enforced in order to create the geometry that would be produced at an impact velocity causing immediate fiber/yarn rupture in transverse impact experiments. The effect of fiber angle around the FSP indenter is experimentally studied along with an analysis of the specific angle causing immediate fiber failure. It is shown that there exists a demonstrative reduction in fiber longitudinal failure strain of KM2 filaments due to this multi-axial stress state, thereby questioning the recurrent assumption that fiber performance within a body armor system is dominated by failure in pure tension.

4.2 Introduction

High performance fibers are utilized in a vast number of applications, such as sporting equipment, turbine fragment containment systems, and of importance to this study, bullet resistant vests. From 1987 to the time of this publication, over 3100
police officers have been saved from death or severe injury due to the implementation of such body armor [115]. Thus, in order to design armor systems that can continue protecting the lives of military and police from advanced threats, it is of extreme importance to fully understand the effect of high-velocity bullet and fragment impact into soft armor systems.

Woven fabric impact is a very complex process, being affected by numerous different parameters. For example, the mechanical response of fabric is a function of the impacting projectile; variables such as striking velocity, mass, nose geometry, and angle of attack all play a role in governing the capability of the fabric to halt the incoming projectile. Furthermore, properties of the fabric such as areal density, weave structure, number of plies, friction interactions, and fiber mechanical properties are all parameters which are known to affect the response of the overall fabric system when placed in a ballistic environment [7,8].

Historically, a large body of work existed before the discovery of the well-known filament Kevlar®, wherein Nylon was utilized in applications such as flack jackets throughout both the Korean and Vietnam wars. In the early investigations, researchers were forced to include non-linear elastic response of fiber into their analysis [20,57,116–119], as compared to the vast majority of more recent work, wherein high-performance fibers can, for the most part, be considered linearly elastic and rate-insensitive [35,89]. This assumption has thereby vastly simplified the understanding of such materials but may have inadvertently caused a slight oversimplification within the research field.

The most basic means of understanding soft armor transverse impact is via shooting a single fiber/yarn with a rigid projectile at its midpoint perpendicular to the fiber longitudinal direction (Figure 4.1). The resulting analysis of the deformation process was developed by a number of authors [117,118,120], but in the US the first major work is attributed to Smith at al. [20], and will be described in greater detail in Section 4.3. Most importantly, these ground laying works implemented the longitudinal tensile failure strain of the fiber as the relevant failure criterion for instantaneous
fiber rupture upon impact. Although a possibly legitimate assumption for Nylon [57], it has been shown that the transverse impact velocity required to promote such an immediate rupture of more modern high-performance yarns is much lower than that predicted from the pure longitudinal rupture strain criterion [50,52,54,121]. Though possibly incorrect, the continuance of such a failure criterion may simply be due to the ease of performing longitudinal tensile tests. It is important to note that Walker and Chocron have developed a plausible explanation for the reduction in this limiting velocity due to fiber bounce, wherein stress waves are emitted inwards from the corners of the projectile, which upon confluence thereby double the stress state present directly in front of the projectile face [54]. Although analytically sound, this treatment of the fiber impact scenario still neglects the effects of any sort of friction or multi-axial loading at the fiber-projectile contact interface.

During the transverse impact of a projectile into a yarn or fiber, the resulting transverse wavefront inherently forces a shear stress to be present within the filament, being typically neglected due to the extremely fine filament diameter (∼10-20 μm). Furthermore, post impact visualization of fibers directly beneath the footprint of the projectile shows an extreme degree of transverse plastic deformation in the form of fiber flattening [11,122]. The question then arises if it is reasonable to neglect such stress states, even if they have historically been brushed aside.

In this light, it is important to note that there does exist a small body of work demonstrating the effect of a multi-axial stress state on resulting fiber mechanical properties, although it has not yet been applied directly towards yarn transverse impact. Several works have found that the torsional modulus of single fibers is directly related to an applied longitudinal stress [76,78,81]. It has also been determined that increasing the torsional strain on single filaments decreases the longitudinal tensile strength [31,55,81,104,123]. Likewise, it has also been recently shown that the tensile strength of the Dyneema® SK76 fiber is affected by the presence of a nominal shear stress, which was demonstrated via the development of a biaxial failure surface criterion [123]. These results are intuitively reasonable as a fiber can eventually be twisted
to failure even without any axial loading. Interesting to note is the lack of longitudinal tensile strength degradation with an applied 1 GPa transverse compression to an 850 denier Kevlar® KM2 fiber with an RC~30 4340 steel indenter [124], presumably due to the lack of damage to intra-filament bonding created in such a loading environment. In contrast, Kevlar® 29 and 49 filaments subjected to a load-release transverse compression demonstrated a marked decrease in post-compression tensile strength, although the failure strain remained unchanged [55]. It is important to note that Kevlar® KM2 and A265 filaments show little reduction in tensile strength due to a similar pre-transverse compression loading history [35,89,125]. Abbott et al. assessed the effect of tying a simple overhand knot in yarns composed of the same Kevlar® 29 and 49 yarns and found a drastic reduction in yarn strength due to a complicated loading state created in the knot [55]. Interestingly, Abbott et al. determined that failure stress trends resulting from progressively twisting both native yarns and yarns with an overhand knot yielded identical failure strength values at elevated twist ratios. The same study found that twisting Kevlar® 29 and 49 yarns to 30 twists per inch (tpi) resulted in a 44% and 48% loss in yarn residual tensile strength, respectively, while single filaments removed from the respective yarns showed a 25% and 32% loss in tensile strength when compared to filaments removed from virgin yarns [55].

In order to evaluate various yarn types with respect to cutting attacks, previous works have been directed at analyzing the cut resistance of yarns [126–128] or single fibers [129]. It was determined that a decrease in fiber longitudinal failure strain exists when subjected to a sharp boundary condition [126]. Mayo and Wetzel showed that single fibers exhibit a large degradation in failure strength when loaded via transverse deflection with a sharp blade [129]. As these works were directed at understanding the cutting phenomenon of high-performance fiber systems, the geometric conditions imposed on the fiber were mimicking the cutting process. To the authors knowledge no previous work has looked more specifically at the stress-state around the projectile geometry at angles relevant to transverse impact. Thus, it is the goal of this work to perform a well-defined experiment that can determine the effect of a local
stress concentration around the corner of a typical test projectile, namely the 0.30 caliber fragment simulation projectile (FSP). In the immediate case, an environment is pursued that is absent from the wave propagation effects described by Walker and Chocron [54] in order to determine if the presence of frictional effects or a multi-axial stress state around the projectile aids in the reduction of the instantaneous rupture velocity. The classical solution to the single fiber/yarn transverse impact scenario will thus be reviewed with a focus on the geometry developed during such an event. The loading geometry in the present experimental setup will then be described, and results will be presented. Finally, the effect of a stress concentration around the projectile corner will be compared with the traditional Smith theory.

4.3 Classical Theory

As previously mentioned, the goal of this work is aimed at re-examining the traditional theory that fibers experiencing the trauma of high-speed transverse impact fail solely via tension, thereby rendering the failure stress in such an environment equal to that of a fiber pulled in pure longitudinal tension within a similar strain-rate regime. Although such an assertion is widespread and the main mechanism of energy dissipation of yarns is via propagation of stress waves along the constituent fibers, it seems an over-simplification to assume a filament is in a constant pure tension environment in the strained portion of the gauge length, especially near the projectile-yarn contact site. This assumption may very well be the culprit causing the drastic over-prediction of the critical velocity at which fibers instantaneously fail when impacted in the transverse direction.

In an effort to simplify the problem to as basic a level as possible, it has been decided that transverse impact into single filaments or into single yarns in a fashion described by Smith et al. [20] will place the problem into an environment which is tractable and will remove the many complexities which can arise during the structural response of an entire fabric system [7,8]. A schematic of this impact orientation
Figure 4.1. Schematic of the transverse impact of a rigid projectile into a yarn or single fiber. Two waveforms are generated, namely, longitudinal and transverse with the former proceeding the latter. Behind the longitudinal wave, \( c \), material flows inward with the particle velocity \( W \). At the transverse wavefront, \( U \), inflowing material changes direction, thereby moving upward with velocity \( V \), identical to the impacting projectile velocity. The developed angle, \( \theta \), remains constant with respect to \( V \) until a reflected wave from the system boundary meets the transverse wavefront.
can be found in Figure 4.1, which represents the impact of a single filament/yarn by a projectile moving in the vertical direction. The resulting wave mechanics will be described similar to the work presented by Smith et al. [20]. Upon contact of the projectile with the horizontal fiber, a tensile stress wave moves outward in the longitudinal direction of the fiber with the wave velocity $c$, which is described in 4.1,

$$c = \sqrt{\frac{E}{\rho}} \quad (4.1)$$

where $E$ and $\rho$ represent the elastic modulus along the fiber direction and the density of the fiber, respectively. Behind the longitudinal wavefront, material is set in motion toward the projectile, and moves with a particle velocity of $W$, being described in 4.2,

$$W = C\epsilon \quad (4.2)$$

where $\epsilon$ represents the strain developed behind the longitudinal wavefront. Upon impact, a second wavefront is also developed which causes the material to form a “tent-like” shape. Defined as the transverse wave, this wavefront moves away from the projectile with velocity $U$, which is described in 4.3,

$$U = \sqrt{\frac{E}{\rho} \left( \frac{\epsilon}{1+\epsilon} \right)^{1/2}} \quad (4.3)$$

Behind the transverse wavefront, the material moves solely upwards in the direction of the projectile with a particle velocity identical to the projectile velocity, $U$, which is described in 4.4.

$$V = \left( (1+\epsilon)^2 U^2 - (1+\epsilon)(U-W)^2 \right)^{1/2} \quad (4.4)$$

It is important to note that the strain behind the longitudinal wavefront, $\epsilon$, is a constant value both in front and behind the transverse wavefront. Finally, the shape
of the wavefront can also be determined via post-process analysis and is defined as $\theta$ and is described in 4.5.

$$\theta = \tan^{-1}\left(\frac{V}{U(1 + \epsilon) - W}\right) \quad (4.5)$$

It is important to note that the deformation geometry described by 4.1 - 4.5 is well substantiated by experimental evidence of previous researchers. Smith et al. impacted a number of different yarn types at low velocities ($\sim$30-50 m/s) and were able to predict the sonic modulus and transverse wave speed with a reasonable level of accuracy [119]. In a later work, Smith et al. impacted high-tenacity Nylon yarn at velocities of roughly 500 m/s and via high-speed flash system were able to show that the transverse wave-speed matched quite well with this theory [57]. Field and Sun impacted both Kevlar® and Spectra fibers under various pretensions, and although not explicitly stated, measured transverse wave speeds agreed extraordinarily well with the Smith equations, if slightly modified to contain a pre-strain value and can be seen in Figure 4.2 [41]. Chocron et al. performed transverse impact tests on Kevlar® KM2, Dyneema® SK65, and PBO yarns, and all yielded experimental transverse wave speeds within $\pm 6\%$ of the theoretical velocity [53]. Although several more works by various authors have shown the success of the Smith equations at predicting the geometry of the transverse impact environment, these will be presently omitted in order to provide brevity. Regardless, the geometry of deformation within the system during impact matches extremely well with the theory presented by Smith et al. [20], thereby giving ample confirmation to 4.1 - 4.5.

As can be seen from Equations 4.1 - 4.5, if one knows the failure strain of a particular fiber material, it is then possible to predict the critical velocity at which the fiber will instantaneously fail upon impact, which for Kevlar® KM2 or Dyneema® SK65 is 945 m/s and 1110 m/s, respectively [53]. Interestingly, when experimentally tested, the actual critical velocity that is measured is much lower, being $\sim$630 m/s and $\sim$550 m/s, respectively, thereby raising the question as to wherein this large variation occurs [53]. As previously stated, the geometry of the deformation process
Figure 4.2. Effect of pretension on the resulting transverse wave velocity during transverse impact of single yarns. Presented data is from Field and Sun (Field and Sun, 1990).

as described by the analytical solution matches extremely well with experimental results below the critical velocity. Furthermore, as aforementioned, if a pretension is placed on the yarn prior to impact, there is a clear increase in the transverse wave velocity, which can be sufficiently predicted with the Smith equations as shown in Figure 4.2, thereby demonstrating that the strain value produced in the wake of the longitudinal wave is well predicted from theory [41]. Thus, because the developed strain in the wake of the longitudinal wave is well described by the Smith solution, it can then be assumed that at the experimental critical velocity, the developed strain in the yarn away from the immediate impact contact site is below that of the actual longitudinal breaking strain of the material.

Currently, the most sound suggestion as to why such a divergence exists between the analytical and experimental critical velocities was produced by Walker and Chocron [54], wherein they postulate that the fiber actually bounces off the surface of the bullet (FSP in their study), as the compression wave which travels through the fiber reflects as a tension wave thereby causing the velocity of the fiber to double, rendering its forward velocity two times that of the incoming projectile. Longitudinal
waves are likewise emanated from the corners in contact with the projectile, moving away in two directions from both corner of the projectile at the impact site. Upon arrival of the longitudinal stress waves that meet directly in front of the projectile, the longitudinal stress state is doubled, which is then attributed as the culprit of the reduced instantaneous rupture velocity. While this elastic theory is rigorously sound and elegant in nature, like the Smith analysis, it also assumes that the fiber always fails in pure tension and that no additional loading effects are created at the projectile-yarn contact site. Furthermore, for such a phenomenon to exist, the projectile face-fiber impact is required to be perfectly parallel in nature. While the possibility of such circumstance is theoretically plausible, with the idea most likely deriving from images sequences of a sphere impacting a rubber band presented by Field and Sun [41], it does not account for any stress concentration that could be present around the indenter corners or the inherently sharp angle produced during the impact event.

Herein lies the culprit of misunderstanding of the impact phenomenon; regardless of what indenter is used for high-speed impact, the same angle $\theta$, from Equation 4.5, is produced, which may play a role in governing the fiber failure mode. For example, it has been shown that transverse impact of UHMWPE fiber by a steel sphere projectile causes a fracture surface dominated by shear failure [50], which is not the typical fibrillated surface seen from this fiber type when loaded to failure in longitudinal tension [38], but rather when cut with a blade [128,130]. Thus, while sharp edged indenters will of course produce more of a stress concentration than rounded indenters, there is still the possibility of a stress concentration at the origin of the transverse wavefront, as a shear stress must be present to have created such an angle $\theta$. For example, work from Prosser et al. has shown that varying the chamfer angle of an indenter head will actually change the failure stress of a yarn if loaded in a transverse quasi static environment [9,11]. Furthermore, it was also shown that different indenter shapes did not have as drastic of an effect in actual transverse impact into a fabric system, which is to be expected, as the angle created
in transverse impact is a function of the impacting velocity, as shown in Equations 4.1 - 4.5 impacting velocity, and the effect of that specific angle on the resulting stress state of the fiber becomes of critical importance, especially at the critical velocity wherein the fiber fails with such a reduced longitudinal tensile strain. Thus, it is the goal of this work to gain insight into the stress concentration present at the corner of an FSP indenter during a quasi-static loading process.

4.4 Experimental Procedure

In order to assess the effect of multi-axial loading on the high performance fiber Kevlar® KM2 without the added complexity of wave mechanics, single filaments are strained via quasi-static transverse deflection loading; a representative image of the system geometry is shown in Figure 4.3. As can be seen in Figure 4.3(a), bollard grips (2 kN pneumatic grips) are used to clamp onto the filament ends. The fiber is loaded via controlled vertical displacement of the indenter using a MTS 810 servo-hydraulic universal testing system. As the cross-head moves the indenter upwards with a constant velocity, the bollard grips holding the fiber ends remain stationary. A schematic of the filament geometry with respect to the indenter and grips can be seen in Figure 4.3(b). Angle $\theta$ is the angle of the fiber with respect to the horizontal axis during the test. For each experiment, the fiber is placed into the apparatus at $\theta_{\text{start}}$ and then loaded until failure, $\theta_{\text{end}}$, using a slack-start procedure similar to that found in ASTM (D7269) standard [112], wherein the cross-head is backed off slightly before the loading process to ensure negligible fiber pretension. A 0.30-cal FSP mild steel projectile is implemented as the indenter, as it is commonly found in transverse impact tests. The FSP utilized is that described by [131] with an increase in chamfer angle to 55°, which allows desired transverse deflection angles to be reached. An image of the FSP indenter along with a measurement of the corner radius of curvature can be seen in Figure 4.4. This measurement was taken via optical microscopy after the indenter was sectioned, mounted in bakelite, and then polished to a mirror finish. It
is important to note that this projectile was machined similar to the traditional FSP geometry, with no effort being made to sharpen the indenter corner. The measured radius of curvature for such an indenter is found to be $\sim 20 \, \mu m$.

In an effort to assess the effect of a possible stress concentration developed around the indenter head at different striking velocities, the starting angle of the filament, angle $\theta_{\text{start}}$ is varied. The variation is performed by either moving the bollard grips horizontally toward or away from the loading piston before the experimental onset, or via altering the indenter initial height with respect to the grips. Altering this start angle inherently enforces varied failure angles, $\theta_{\text{fail}}$. Four different $\theta_{\text{start}}$ values are utilized, thereby enforcing a diverse range of failure angles. It is also important to
Figure 4.4. (a) Modified 0.30 caliber FSP used as the indenter head. (b) Polished corner of the FSP shows a radius of curvature of $\sim 20 \mu m$. 
note that due to the low failure strain of the tested materials (∼3-4% strain), \( \theta_{\text{fail}} \) is at most a few degrees greater than \( \theta_{\text{start}} \), with this difference decreasing for larger \( \theta_{\text{start}} \).

A load cell (Interface 1500ASK-25) located beneath the indenter is used to track the position of fiber failure and to help ensure that only a single fiber is tested. The load signal is appropriately amplified by a Vishay 2310B signal conditioner. The displacement of the cross-head is determined via direct measurement of an output signal provided by the MTS 810 controller. Both the amplified load and displacement signals are collected with a dual channel Tektronix DPO4032 oscilloscope at a sampling frequency of 250 Hz. Cross-head velocity is set at 1.2 mm/sec for all tests in order to ensure negligible vibrations induced to the load cell, resulting in a nominal strain rate of 0.005/s.

In order to ensure negligible slipping, carbon tape (Tad Pella 16073-1) is affixed to the gripping platens for all experiments, as it functions surprisingly well at holding single fibers when combined with transverse compressive gripping. Furthermore, fibers in various experiments are flagged near the end of the gripping zone in order to track the appearance of grip slip; in all cases tested, fiber slipping appears non-existent. This is also corroborated by the lack of sudden drops in the load signal, which is a common load-response for poor adhesive bonding in single fiber testing. An example of the load signal as a function of time can be seen in Figure 4.5.

Using the output crosshead displacement signal and the geometry of the system, the total fiber length between the two bollard grips can be determined as a function of time. At the time of failure, which is demarcated by a drop in the load signal to zero, the length of the fiber is determined \( L_{\text{end}} \), thereby allowing for measurement of the longitudinal failure strain exhibited by the fiber, which is defined as,

\[
\epsilon_{\text{fail}} = \frac{L_{\text{end}} - L_{\text{start}}}{L_{\text{start}}} \tag{4.6}
\]

\( L_{\text{start}} \) represents the single filament length at the start of the experiment, which is determined using a computer aided design software via inputting the crosshead
location output from the MTS. $L_{\text{start}}$ varies slightly depending on $\theta_{\text{start}}$, but a nominal length of roughly 550 mm is implemented for the various geometric conditions. Furthermore, the angle at failure is determined via post-process analysis using known geometry conditions. As an example of crosshead displacement and measured load versus time, Figure 4.5(a) shows the results from a single fiber being loaded at a $\sim 50^\circ$ angle using the FSP indenter. Figure 4.5(b) demonstrates the difference in fiber angle from $\theta_{\text{start}}$ to $\theta_{\text{fail}}$ as a function of $\theta_{\text{fail}}$. This reduction in angle range decreases with increasing $\theta_{\text{fail}}$ due to the experimental geometry.

To possess a representative baseline failure strain, single fibers are also pulled in longitudinal tension, using a hybrid approach between ASTM (D7269) and ASTM (1557-03), representing yarn and single fiber quasi-static tension tests, respectively [112, 114]. For the three shorter gauge lengths, the traditional tabbing method is employed (ASTM 1557-03) and for the longest gauge length bollard grips are utilized (ASTM D7269), similar to longitudinal yarn tests, so as to create a comparable boundary condition to that provided in the transverse displacement experiments. Furthermore, this set of experiments is used to additionally ensure that fiber slippage is not present at the bollard-clamping site. The slack-start procedure is implemented for these longitudinal tension tests and all reported values are derived from samples failing within the gauge length [112].

4.5 Results and Discussion

The high performance fiber Kevlar® KM2 is subjected to transverse deflection loading as previously shown in Figure 4.3. The angle $\theta_{\text{start}}$ is varied to force various $\theta_{\text{fail}}$, thereby allowing for an experimental analysis of the effect of $\theta_{\text{fail}}$ on the resulting longitudinal failure strain of the fiber. The angle $\theta_{\text{fail}}$ is varied between $0^\circ$ and $50^\circ$ and the resulting longitudinal failure strain values are plotted in Figure 4.6(a). As can be seen, there is a clear drop in the longitudinal failure strain with increasing angle, $\theta_{\text{fail}}$. It is important to note that all observed failures of the fiber occur at the FSP
Figure 4.5. (a) An example of the crosshead displacement and vertical force. This specific experiment consists of a Kevlar® KM2 fiber loaded at \(\sim 50^\circ\), using the FSP indenter. (b) The range from starting angle \((\theta_{\text{start}})\) to failure angle \((\theta_{\text{fail}})\) as shown in Figure 4.3. The angle change decreases with increasing failure angle due to the increased starting angle.
corner, with verification performed via inspection of the filament failure location with respect to the indenter proximity post-mortem. In this scenario, it is also important to reiterate that there is no effect of fiber bouncing or wave propagation, as the material is being loaded in a quasi-static environment, yet there is still a stark reduction in failure strain with increasing angle. If the classical theory that the fiber fails in pure tension no matter the loading condition, there would be no drop in the failure strain. In efforts to ensure the data processing is correct, load data from the force transducer located directly beneath the indenter has been recorded at failure and the resulting stress along the axial direction of the fiber has been calculated, which is shown in Figure 4.6(b). As expected, the reduction in failure strain shown Figure 4.6(a) and failure stress shown in Figure 4.6(b) are quite similar, due to the linear elastic nature of the Kevlar® KM2 fiber. Data from Figure 4.6(a) has also been reorganized and plotted in Figure 4.6(c) along with previous data from Prosser, who rather than changing the angle of the fiber around a fixed FSP geometry, altered the chamfer angle of the loading geometry [9]. The Prosser data was originally presented as failure stress, but assuming a linear elastic stress-strain curve for Kevlar®, these results are normalized to their longitudinal failure stress values and overlaid on this plot. Clearly, there is a demonstrative effect of failure angle on the resulting longitudinal failure strain in the absence of any wave mechanics or fiber bouncing. It is also important to note that other authors have seen this reduction in failure strain when loading fibers in an off axis loading-scheme, namely Shockey et al. found a decrease in failure strain for both Kevlar® and Zylon yarns when loaded with a razor blade via transverse deflection, albeit the loading angles were quite small, being roughly 10° [126].

As the failure strain value measured in the pure longitudinal zero degree experiments is less than that reported by Walker and Chocron [54], a gauge length effect study is performed. Sample lengths of 5 mm, 10 mm, and 50 mm are pulled in tension at a strain rate of 0.001/s using the traditional tab gripping method described in ASTM standard D1557-03, and the resulting failure strains are shown in Figure 4.6, along with data resulting from pure tension tests using the bollards grip and a gauge
Figure 4.6. (a) Breaking strain of a single Kevlar® KM2 fiber as a function of the failure angle. There is a clear decrease in failure strain with an increasing failure angle, $\theta_{\text{fail}}$. (b) Breaking stress developed along the longitudinal direction of the fiber at failure as processed from the vertical load sensor. (c) Strain data normalized to the longitudinal failure strain average and then re-plotted along with data from Prosser [9].
Figure 4.7. In order to check the validity of the measured failure at 0\(^\circ\) strain in the previously listed data of Figure 4.6, a gauge length effect study was performed.

length of 557 mm. As can be seen, there is a definite drop in the failure strain for increasing gauge lengths. Thus, the pure tension failure strain value shown in Figure 4.6, although lower than the typically listed value of \(\sim 4\%\), is deemed reasonable. From the experimental results in Figure 4.6, it also seems fair to assume that there is fiber defect possessing a periodicity between 10-50 mm that greatly reduces the fiber failure strain in pure tension. As the fibers loaded in the transverse deflection environment consistently fail around the corner of the FSP indenter, it is speculated that a localized region of stress concentration is developed in the fiber at the contact site. Thus, it is assumed that the aforementioned defining defect in the length scale of 10-50 mm can be neglected as failure occurs at the FSP corner, not at a random site within gauge length. Furthermore, at increased \(\theta_{\text{fail}}\), the failure strain continues to decrease, which would not occur if failure were caused by such a defect.

It is important to note that the experimental results in Figure 4.6 depict the reduction in failure strain \(\varepsilon_{\text{fail}}^{\text{quasi}}\) as a function of breaking angle \(\theta_{\text{fail}}^{\text{quasi}}\), which is inherently forced by varying the initial starting angle, \(\theta_{\text{start}}^{\text{quasi}}\). Thus, as previously stated, this evolution purely shows the decrease in failure strain as a function of rupture
Table 4.1. Resulting mechanical properties from single fiber tension experiments following ASTM standard D1557-03. Compliance correction as been performed on gauge lengths of 5, 10, and 50 mm as the tab gripping method was used said samples.

<table>
<thead>
<tr>
<th>Gauge Length (mm)</th>
<th>Failure Strain (%)</th>
<th>Failure Stress (GPa)</th>
<th>Elastic Modulus (GPa)</th>
<th>Number of Samples</th>
<th>Compliance Corrected Failure Strain (%)</th>
<th>Compliance Corrected Modulus (GPa)</th>
</tr>
</thead>
<tbody>
<tr>
<td>5</td>
<td>4.68±0.61</td>
<td>3.79±0.33</td>
<td>81.68±8.83</td>
<td>15</td>
<td>4.01</td>
<td>94.22</td>
</tr>
<tr>
<td>10</td>
<td>4.18±0.45</td>
<td>3.63±0.34</td>
<td>87.27±6.88</td>
<td>30</td>
<td>3.86</td>
<td>94.22</td>
</tr>
<tr>
<td>50</td>
<td>3.28±0.64</td>
<td>3.04±0.68</td>
<td>92.56±6.23</td>
<td>30</td>
<td>3.23</td>
<td>94.22</td>
</tr>
<tr>
<td>557</td>
<td>3.03±0.32</td>
<td>3.06±0.26</td>
<td>98.69±5.51</td>
<td>30</td>
<td>N/A</td>
<td>N/A</td>
</tr>
</tbody>
</table>
angle, \( \theta_{\text{fail}}^{\text{quasi}} \). Clearly the fiber is not failing in pure tension at elevated angles, \( \theta_{\text{fail}}^{\text{quasi}} \). Although this finding is a critical point, what becomes even more interesting is the analysis of the failure strain at the angle for which would be generated when the filament/yarn is impacted at a velocity wherein instantaneous rupture occurs. This angle can be found by solving for \( \theta_{\text{fail}}^{\text{impact}} \) in Equations 4.1 - 4.5 using the critical velocity, \( V_{\text{fail}}^{\text{Chocron}} \) of 627 m/s given by Chocron et al. [53]. Furthermore, at this specific velocity, \( V_{\text{fail}}^{\text{Chocron}} \), not only is the \( \theta_{\text{fail}}^{\text{impact}} \) value known, but \( \varepsilon_{\text{fail}}^{\text{impact}} \) is also defined via solving Equations 4.1 - 4.5, which as aforementioned, is below the failure strain of the fiber in pure longitudinal tension. Interestingly, the strain \( \varepsilon_{\text{fail}}^{\text{impact}} \) developed by impact at this \( V_{\text{fail}}^{\text{Chocron}} \) falls remarkably close to the failure strain (\( \varepsilon_{\text{fail}}^{\text{quasi}} \)) of the fiber in the quasi-static transverse loading experiments at the angle that would be produced in the yarn when impacted at this \( V_{\text{fail}}^{\text{Chocron}} \), which is shown in Figure 4.8(a).

For the Kevlar® KM2 fiber discussed thus far, the failure strain (\( \varepsilon_{\text{fail}}^{\text{impact}} \)) of the fiber when impacted at the critical velocity (\( V_{\text{fail}}^{\text{Chocron}} \)) is determined by the Smith equations to be 0.0211, with a theorized angle (\( \theta_{\text{fail}}^{\text{impact}} \)) of 31.1°, using a modulus of 98.7 GPa determined from the zero degree pure tension tests. As shown in Figure 4.8, the Smith solution intersects the quasi-static transverse loading experimental curve at a failure angle (\( \theta_{\text{start}}^{\text{quasi}} \)) of 31.9° and failure strain (\( \varepsilon_{\text{fail}}^{\text{quasi}} \)) 0.0237, which only differs \( \sim10\% \) transverse impact test results presented by Chocron et al. [53]. If the modulus used to determine the resulting angle and strain values is 79.9 GPa given by Chocron et al. (Chocron, et al. 2011), the resulting angle (\( \theta_{\text{fail}}^{\text{impact}} \)) and strain (\( \varepsilon_{\text{fail}}^{\text{impact}} \)) at are 32.234° and 0.0243, respectively. This failure strain (\( \theta_{\text{fail}}^{\text{impact}} \)) is even closer to that predicted from the current data set, differing above by only \( \sim3\% \).

At this point it is important to reiterate the meaning of each curve shown in Figure 4.8. The current data set curve represents a fit of the data shown in Figure 4.6(a) where at a specific angle of loading the fiber ultimately fails with a corresponding strain value shown in the plot. In contrast, the Smith relations give the corresponding strain that would be developed behind the longitudinal wavefront when the fiber is impacted by a velocity promoting a transverse wave with angle \( \theta_{\text{impact}} \). Thus the
Smith solution is not describing a specific failure angle, rather it solely gives the geometry and strains of the resulting deformation process. The idea of a critical velocity comes from the fact that at some velocity $V_{critical}$, upon contact with the projectile, the fiber immediately fails as it has reached a critical failure strain, which is generally believed to be identical to the failure strain determined from longitudinal tensile tests. Herein lies the heart of the problem, at the critical velocity measured in experiments, Equations 4.1 - 4.5 describe a developed strain value much less than the failure strain from longitudinal tension tests. It must be reiterated that the geometry of the Smith relations predicts remarkably well the shape of the resulting transverse wavefront as a function of time [53,57], the speed of the longitudinal wavefront [41], and the effect of prestress on the resulting transverse wavefront velocity [41], thereby providing adequate validity to Equations 4.1 - 4.5. It is thus postulated that such equations can be used to justifiably determine angle ($\theta_{\text{fail}}^{impact}$) and strain ($\varepsilon_{\text{fail}}^{impact}$) at the critical velocity $V_{critical}$. Using the critical velocity ($V_{Chocron,\text{Chocron}}^{critical}$) given by Chocron et al. [53], the resulting angle ($\theta_{\text{fail}}^{impact}$) and strain ($\varepsilon_{\text{fail}}^{impact}$) are determined and are demarcated by the star symbol in Figure 4.8(a). Again, what is quite interesting is that data from the current quasi-static tests fall remarkably close to the angle ($\theta_{\text{fail}}^{impact}$) and strain ($\varepsilon_{\text{fail}}^{impact}$) conditions predicted by the Smith relations.

In order to further visualize the drastic effect of the loading angle on the resulting failure strain of the fiber, data from the 10 mm longitudinal tests shown in Figure 4.7 has been re-plotted in Figure 4.8(b). As previously mentioned, because failure occurs at the corner geometry of the FSP indenter, it is reasonable to assume that the region around the indenter affected by the stress concentration is less than 10 mm and the corresponding 10 mm failure strain of the fiber at 0o is a more appropriate means of analyzing the effect of the FSP corner geometry. Thus, the data fit is forced through a failure strain value of 4%.

As shown in Figure 4.8, clearly the effect of a multi-axial stress or at least a more complex stress state must be taken into account during the transverse impact of a single fiber/yarn, as this simple approach of loading the fiber quasi-statically into a
Figure 4.8. Experimental results of the Kevlar® KM2 fiber longitudinal failure strain as a function of the failure angle is plotted in grey along with the theoretical angle $\theta_{\text{fail}}^{\text{impact}}$ and strain $(\varepsilon_{\text{fail}}^{\text{impact}})$ generated during transverse impact as predicted from the Smith equations, which is represented by the black curve. While (a) presents the results with the 0° failure strain using the bollard grip assembly which has a gauge length of 557 mm, (b) presents the 0° data using the 10 mm gauge length sample data. Finally, it is important to note that the velocity scale on the right-hand ordinates of (a) and (b) is not equally spaced so as to match up with the linearly scaled strain values on the left hand ordinates.
similar geometry to that seen in transverse impact results in a demonstrative decrease in the critical theoretical transverse impact velocity. Strikingly, the location at which the quasi-static test results intersect with the Smith solution is quite close to the strain ($\varepsilon^{\text{impact}}_{\text{fail}}$) and angle ($\theta^{\text{impact}}_{\text{fail}}$) created when impacting at the experimental breaking velocity ($V_{\text{Chocron critical}}^{\text{Chocron}}$) reported by Chocron et al. [53]. As previously stated, the indenter used in this study has a measured radius of curvature of 20 $\mu$m, which is speculated to cause the resulting loss in failure strain due to a stress concentration developed around the indenter corner. It is important to note that although the FSP geometry has been chosen due to its use in studies of open literature, other penetrator geometries in literature have been implemented to find yarn critical velocities, most notably in experiments performed by Carr using 5.5 mm spheres [50]. Testing both Kevlar® (129 and KM2) and Dyneema® (SK66) fibers, it was determined that a reduction in critical velocity is indeed seen when using the 5.5 mm penetrator, which possesses a radius of curvature two decades larger than the current quasi-static experiments. At first glance this may prove unsettling, but in actual transverse impact experiments it can be postulated that fibers may experience a sharp stress concentration at the kink formed by the ensuing transverse wavefront. Furthermore, for such a transverse wavefront to be developed, it is inherently required that a shear stress be present upon the onset of its development, which is the time period in which yarns shot at or above the critical velocity immediately fail. It is quite possible that the radius of curvature during such a development very well may be on the order of the fiber diameter.

Above the ballistic limit of a fabric, failure is reported to occur locally in close proximity to the impacting projectile footprint [7, 9, 126]. Below this critical velocity, energy dissipation mechanisms such as fiber friction and yarn uncrimp become of importance [8], yet the much more intriguing aspect of the impact phenomenon is when the projectile velocity is close to or above the critical velocity. In this circumstance, the fibers around the projectile fail almost instantaneously and as reported by Cunniff, exhibit an inelastic response [49] due to the lack of longitudinal stress
wave propagation. This immediate failure is attributed to the presence of a transverse compression and shear environment locally around the projectile [7].

Furthermore, close to the $V_{50}$ velocity, because the first several layers of the fabric behave inelastically, Cunniff also showed that initial fabric plys could be replaced with a sacrificial material that is less costly than the high-performance fabric, achieving a comparable or even better ballistic limit [49]. The origin of this inelastic behavior felt by the first several layers of fabric can quite possibly be explained by the current data set. This initial material must take the brunt of the high stress concentration caused by the indenter corner geometry coupled with the angle produced by the ensuing transverse wave. As previously shown in Figure 4.6, there is a definite drop in the fiber failure strain produced by the stress concentration around the projectile corner. Thus, the initial layers of fabric possess a greatly attenuated resistance to failure when compared to their longitudinal failure strain. At a low enough velocity, material punched out by these first few layers, although failing quite rapidly, are then able to effectively smooth out the resulting stress concentration for the remaining high-performance fabric layers. Of course if the velocity is high enough, the entire fabric system will behave inelastically and will be completely penetrated without reasonable energy dissipation, which can be seen in typical residual-velocity versus strike-velocity experimental data [7].

Finally it is again important to reiterate that if fibers/yarns fail almost immediately upon impact, they will not sufficiently move energy away from the impact site and their dissipation capability is thereby voided. The obvious goal in soft body armor is to engage as many yarns as possible and to ensure that they are able to transmit strain energy along their length before rupture. If this is not achieved and the projectile is able to penetrate through a layer of fabric, then said layer becomes essentially useless for the remainder of the impact event. Thus, it is of great importance to understand if the presence of a multi-axial stress-state does exist how it produces failure in a fiber/yarn. Accordingly, the assumption of implementing solely
the longitudinal ultimate strain or stress into a failure criterion for high-performance fiber systems may be lacking.

4.6 Conclusions

The high performance fiber, Kevlar® KM2, is loaded in a quasi-static transverse deflection environment in efforts to reveal the effect of the FSP corner geometry caused by the transverse wave during normal impact of a single fiber/yarn. It is shown that a drastic effect of angle on the resulting longitudinal failure strain of the material exists due to a local stress concentration developed at the FSP corner. Resulting failure strain values are compared with actual transverse impact data from Kevlar® KM2 presented by Chocron et al. [53], and a surprisingly effective correlation exists between the current data set and the experimental transverse impact velocity that causes instantaneous fiber/yarn rupture. It is thus suggested that the age-old merit parameter of high-performance fiber systems be redefined in efforts to also include the effect of the stress concentration that occurs around the projectile due to the inherent transverse wavefront that is developed.
5. The effects of off-axis loading on the failure strain of various high-performance fibers

Adapted from:

5.1 Abstract

Single filaments are subjected to a transverse deflection loading environment in efforts to gain insight into the failure strain of soft-body armor systems experiencing transverse impact. The fiber types utilized for all such experiments are Kevlar® KM2, Spectra® 130d, Dyneema® SK62, Dyneema® SK76, and Zylon® 555. In order to understand the effect of indenter shape, three different indenter geometries are utilized, namely a 0.30 caliber rounded head, a 0.30 caliber fragment simulation projectile (FSP), and high-carbon steel razor blades. The angle at failure is also varied in order evaluate the presence of a stress concentration developed around such indenters through angles that would be produced during the transverse impact of single fibers/yarns. Loading with the rounded indenter yields failure strain values similar to pure longitudinal tensile experiments. Fibers loaded via razor blade show a drastic reduction in failure strain, although demonstrated failure strains are reasonably similar for all tested angles. Most interestingly, fibers loaded with the FSP show a reduction in failure strain with increasing loading angles, with low angle and high angle failure strains being similar to failure strains of fibers loaded with the rounded indenter and razor blade, respectively. In efforts to gain further insight into the method of fiber failure due to different loading configurations, post-mortem fracture surfaces are imaged for Kevlar® KM2 and Dyneema® SK76.
5.2 Introduction

High performance fibers are often included in advanced structural systems due to their profound tensile strength to density ratio. Due to ease, the vast majority of mechanical testing on such filaments and yarns is in the form of longitudinal tensile testing, yielding a linear-elastic stress-strain response for many fiber types [21,22,28,37,132]. This is due to the extremely high level of orientation present for polymeric fibers, resulting in crystallinity values of 75-95% [22,28]. Coupled with their low density (∼1,000-2,000 kg/m³), they are most routinely employed in products requiring high strength and high stiffness, namely soft or hard composite structures including, hulls of military and sporting watercraft, mooring lines, anti-spall linings in armored vehicles, turbine fragment containment barriers, and of most interest to the current study, body armor. With regards to the latter application, understanding behavior of the constituent high performance fibers in both ballistic impact and cutting environments is of extreme importance.

Currently, the most effective means of determining the halting capability of a woven fabric is by actual impact and cutting experiments, which are of course an extremely cost prohibitive and time consuming step in the product evaluation process, as large amounts of material are needed to weave a ballistic panel. That said, in recent years much effort has been placed on developing reliable numerical models in efforts to predict how a certain vest may perform, with the capability of altering parameters such as fiber type, weave structure, and the number of plies. Although full system modeling is of great use and is an extremely powerful technique, very few works have adequately achieved predictions of fabric performance. To the authors’ knowledge, the most mature constitutive modeling efforts are currently seen from Southwest Research Institute (SwRI), which are capable of adequately predicting ballistic limit velocities for full fabric systems [133,134].

Although these computational methods are quite insightful, additional works have shown that non-dimensional analysis can be of use to understand ballistic impact of
soft-armor systems. Cunniff developed a non-dimensional parameter that contains the most relevant physical properties involved during transverse impact, namely longitudinal wave speed, $\sqrt{\frac{E}{\rho}}$, and specific longitudinal yarn toughness, $\frac{\sigma_e}{2\rho}$ [10]. The Cunniff parameter is thus described by the following:

$$U^* = \frac{\sigma_e}{2\rho} \sqrt{\frac{E}{\rho}} \quad (5.1)$$

where $\sigma$, $\rho$, and $E$ represent the fiber tensile strength, failure strain, density, and longitudinal modulus, respectively. The essence of the parameter $U$ represents the energy density required to rupture the fiber material ($\frac{\sigma_e}{2\rho}$) and the velocity at which energy can be moved away from the impact site, $\sqrt{\frac{E}{\rho}}$. In practice, $U^{1/3}$ is used to normalize the fabric ballistic limit ($V_{50}$) data from different armor systems, and if effective, can collapse the response from different material systems down onto one master curve when plotted against $\frac{A_d A_p}{m_p}$, where $A_d$, $A_p$, and $m_p$ represent system areal density, projectile presented area, and projectile mass, respectively. Phoenix and Porwal furthered the power of this solution by assuming that the impacted fabric can be modeled as a membrane [135]. Using work originally developed by Rakhmatulin [116], they were thereby able to ascertain a closed form solution identical to that from Cunniff while also uncovering additional physical insight into the phenomenon such as a solution describing the necessary distance needed to halt the projectile.

It is important to remember the two key physical parameters found in Equation 5.1, namely, the specific yarn toughness and fiber longitudinal wave speed; the vast majority of impact energy is dissipated via transportation of strain energy away from the impact site at the inherent wave speed in the material. Therefore increased yarn specific toughness allows for more energy absorption per unit of material, while increased wave speed moves such energy away from the impact site at a higher rate. Thus, a reduction in either of these values will inherently reduce the effectiveness of a soft armor system.
The relevance of both longitudinal wave speed and toughness within the parameter is corroborated by external modeling efforts of Roylance and Wang who have shown that the majority of energy during an impact event is dissipated via longitudinal strain energy, longitudinal kinetic energy due to yarn movement from the longitudinal wave, and out-of-plane kinetic energy initiated via passage of the transverse wave front [136]. With this understanding it then becomes of interest to examine if the longitudinal toughness is an unchanged value, especially when the fiber region around the projectile face is loaded in a complex environment that may force the filament to undergo either multi-axial loading conditions, an increase in tensile and compressive strains due to bending around the indenter head, or frictional effects between the fiber surface and projectile. Furthermore, each of the aforementioned works assumes an unaltered failure criterion regardless of loading condition. Such a failure mode is the typical assumption made throughout literature, developing from the classical Nylon work that was performed by Smith, et al. [20, 57]. Although a reasonable approximation at the time of the Smith, et al. studies, more recent work has shown the effect of multi-axial loading on single fibers, fabrics, and yarns. In fact, it has even been stated that a multi-axial stress state must exist in the fibers directly beneath the footprint of the projectile during transverse impact [7, 9, 49, 127, 137, 138], although to the authors’ knowledge such assertions have not altered the failure criterion used in modeling efforts.

In high velocity transverse impact of fabric or composite systems, the vast majority of fiber/yarn failure occurs directly beneath the projectile [9, 11, 49, 126]. Thus, the ineptitude of various fabric performance parameters or failure criterion may reside in the absence of a relevant failure criterion for the single filament. It has been shown that an imposed torsional shear strain alters the resulting longitudinal tensile strength of various high-performance fibers [31, 55, 104, 123]. Furthermore, Abbott et al. [55] found that filaments removed from Kevlar® 29 and 49 yarns undergoing 30 twists per inch yielded longitudinal failure strength values 25% and 32% below strength values from filaments extracted from untwisted yarns. It was also determined that
Kevlar® 29 and 49 single filaments subjected to 20 N/mm transverse compression exhibited a 22-25% reduction in residual tensile strength [55]. In contrast, previous work on Kevlar KM2® and A265 filaments has shown minimal longitudinal tensile stress-strain degradation to fibers subjected to transverse compression [35,125] [36]; Lim, 2010; [124]. More closely related to loading seen in body armor materials, it has been shown that transverse deflection of single yarns can degrade the resulting longitudinal failure strain when loaded in a transverse direction with a sharp indenter [126,127,138]. It has also been reported that either increasing the blade sharpness or yarn pretension decreases the resulting failure strain of the yarn when loaded in a transverse deflection environment [128]. When loading single fibers with a transverse cutting mechanism, Mayo and Wetzel demonstrated that the highly anisotropic nature of organic fibers governs failure, and it is alluded that a shear failure criterion may be more realistic to predict failure rather than solely tension, as on average, the aramid and UHMWPE (ultra-high molecular weight polyethylene) fibers tested ruptured at stress values ranging from 22-48% of their longitudinal tensile failure strength [129]. On the fabric level, it has also been shown that a sharp blade more easily pushes through a woven ply as compared to a blunt blade, with the majority of the fibers failing at the indenter contact site [126].

Although the previous list of works demonstrates the presence of a multi-axial stress state of the constituent fibers during a cutting environment, of chief concern to the present work is the possibility of a multi-axial loading condition experienced by a fiber around a projectile during transverse impact. With regards to actual transverse impact of single yarns, it has been consistently shown that the velocity required to promote near-immediate rupture upon projectile-yarn contact is much lower than that predicted when using the longitudinal rupture strain as the relevant failure criterion [50,52,54,55,121]. In a previous study, the present authors found that transverse deflection of a single Kevlar KM2 fiber with a fragment simulation projectile (FSP) caused a stark reduction in the resulting single filament failure strain when loaded at geometries similar to that produced during single fiber/yarn transverse
impact (Hudspeth, submitted). Such a quasi-static environment was utilized in order to negate the effects of wave mechanics, which has been attributed to be the culprit of the reduced instantaneous failure velocity of a yarn subjected to transverse impact [54].

The aforementioned pure tension assumption continues to be presented throughout literature, from single fiber transverse impact analysis, to the vast majority of constitutive models of full-fabric. To the authors’ knowledge, no published literature exists that investigates the possibility of a multi-axial stress-state or curvature effects within the geometry present during transverse impact, although the possibility of its importance is mentioned on various occasions [7, 9, 49, 127, 137, 138]. Albeit, SwRI does utilize an orthotropic material model in their transverse impact simulations [133, 134], but the implemented failure model is a von Mises surface that is applicable to an isotropic material, which is not representative of high-performance fibers [35, 36, 89, 125].

Thus, it is proposed that the fidelity of fabric modeling tools may be increased if they contain the proper multi-axial (out-of-plane) stress considerations. In light of this necessity, the aim of the ensuing work is to develop a parametric set of experimental results showing the effect of projectile nose geometry and fiber breaking angle. The former condition is probed in order to determine if projectile geometry has any effect on local stress concentrations exhibited by various fiber types. The latter condition is analyzed in order to determine if the geometry condition that is developed by the transverse wave front in a soft-armor impact event plays any role in reducing the ultimate energy absorbing capability of the fabric. Finally, filament rupture surfaces generated from the varying test angles and indenters are imaged from each fiber class, namely Dyneema® SK76 and Kevlar® KM2, which are archetype representatives for UHMWPEs and aramids, respectively.
Table 5.1. Fiber properties collected from producer data sheets and previous testing. Note: Kevlar® KM2 and Dyneema® SK76 properties are measured from single filament tests, while Dyneema® SK62, Zylon® AS-555, and Spectra® 130D properties are presumably measured by the manufacturer via yarn testing. Diameter measurements are taken from as-received samples for Dyneema® SK62, Zylon® 555 and Spectra® 130D.

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<tr>
<td>Density (kg/m³)</td>
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<td>980</td>
<td>970</td>
<td>1540</td>
<td>970</td>
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<tr>
<td>Young’s Modulus (GPa)</td>
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<td>132</td>
<td>65-100</td>
<td>180</td>
<td>113</td>
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<tr>
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<td>2.4-3.3</td>
<td>5.8</td>
<td>3.25</td>
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<tr>
<td>Failure Strain, Reported (%)</td>
<td>4.52±0.37</td>
<td>3.5</td>
<td>3-4</td>
<td>3.5</td>
<td>3.2</td>
</tr>
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<td>Failure Strain, Current Data (%)</td>
<td>3.03±0.32</td>
<td>3.54±0.32</td>
<td>3.51±0.57</td>
<td>3.00±0.41</td>
<td>3.56±0.30</td>
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<tr>
<td>Fiber Diameter (µm)</td>
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<td>16.1</td>
<td>19.8±1.6</td>
<td>11.4±0.7</td>
<td>24.4±4.5</td>
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<tr>
<td>Yarn Linear Density (denier)</td>
<td>600</td>
<td>1350</td>
<td>2400</td>
<td>500</td>
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5.3 Experimental Procedure

Several different organic fiber types are loaded in a transverse deflection environment containing a similar geometry to that produced during transverse impact in order to determine if the effect of a multi-axial loading condition or a curvature effect around the indenter head is relevant. As aforementioned, previous work by the present authors demonstrated the deleterious effect of loading single Kevlar® KM2 fibers with a fragment simulation projectile (FSP) indenter at various deflection angles [139]. Thus, in order to continue this work to additional fiber types, several different single filaments are tested, namely, Kevlar® KM2, Spectra® 130d, Dyneema® SK62, Dyneema® SK76, and Zylon® AS-555. Longitudinal tensile properties of each fiber type are presented in Table 5.1 with most results purposefully coming from manufacturer data sheets if readily available.

As the experimental geometry currently utilized is similar to that found in Hudspeth et al. (Hudspeth et al., submitted), only a short description of the experimental setup is provided. Filaments are first carefully removed from fiber bundles via isolating a single fiber from the yarn and then sliding it out lengthwise along the direction of the tow. Said single filaments are then placed into the loading device illustrated.
in Figure 5.1, which is capable of producing various deflection angles with a nominal fiber gauge length of 0.55 m. In this setup, in order to alter the fiber failure angle, the starting angle of the fiber is accordingly changed. Altering is performed by either changing the starting height of the indenter or by moving the gripping mechanism of the fiber inward or outward. The gripping method utilized is via bollards that are traditionally used in longitudinal yarn testing. In efforts to negate slippage of the fiber between the gripping platens, carbon tape (Tad Pella 16073-1) is mounted on the platen assembly. Flags are affixed to the ends of the fiber near the bollards and subsequent tracking reveals no apparent slippage of the fiber between the grips. Furthermore, the load signal generated from the attached load cell reveals no sudden force drops, which if present, would be indicative of fiber slip. Although it may be postulated that slipping could occur without such apparent load drops, the lack of discernible movement of the aforementioned flags along with the measured strain being comparable to traditional fiber failure strains, leads the authors to believe that such slipping, if present, is negligible.

In efforts to determine if there is an effect of curvature radius around the indenter corner, three different indenter heads are implemented; the 0.30 caliber fragment simulation projectile (FSP), a razor blade, and a rounded indenter are utilized, with each of these being presented in Figure 5.2. Furthermore, the radius of curvature from each indenter head is measured and can be found in Figure 5.3. For the FSP and razor blade indenters, measurement is performed by sectioning the indenter, polishing the cross-section to a mirror finish, and then imaging the corner geometry via optical microscopy. The radius of curvature of the round indenter does need to be directly measured via microscopy, as it is of appropriate size to be determined via calipers. The radii of curvature for the round, FSP, and razor blade indenters are 3.8 mm, \( \sim 20 \, \mu m \), and \( \sim 2.3 \, \mu m \), respectively. Furthermore, the design of the FSP follows dimensions given by MIL-DTL-46593B, but with an increase in chamfer angle to 55o that allows for high fiber angles to be achieved [131]. The razor blade is typical high-carbon steel (Personna 94-0152) and is changed for each experiment.
Figure 5.1. (a) Image and (b) representative schematic of the experimental setup used to load single fibers via transverse deflection. Note $\theta_{\text{start}}$ is strategically varied so as to produce specific $\theta_{\text{fail}}$. Also of note in (a) is the use of a yarn instead of a single fiber in order to provide adequate visualization.
Figure 5.2. Three different indenters used for transverse deflection loading. From left to right they are a 0.30-caliber FSP, a 0.30-caliber rounded head, and a razor blade. It is important to note that the razor blade was changed for all tests in efforts to minimize blade dulling.
Figure 5.3. Radius of curvature of both the (a) FSP and (b) razor blade is measured via polishing the indenter cross-section.
Indenter displacement is determined via measurement of a signal output from a servo-hydraulic load frame (MTS 810) while the vertical load produced onto the indenter is sensed with a force transducer (Interface 1500ASK-25). Both signals are simultaneously tracked on a Tektronix DPO4032 oscilloscope at a sampling frequency of 250 Hz, and a representative set of results can be seen in Figure 5.4(a). The resulting displacement data is then used to determine the failure strain of the fiber via post-process analysis using a computer aided design software and known geometry conditions of the experimental setup. Furthermore, the starting and failure angles of the fiber are determined, and due to the low failure strain exhibited by these fiber types, both of these angles only differed by a few degrees. The starting angle, $\theta_{\text{start}}$, is recorded just as a load is measured from the force transducer, thus at a zero force pretension of the fiber, while the failure angle, $\theta_{\text{fail}}$, is recorded when the load measured by the transducer suddenly drops to zero. An example of the range between $\theta_{\text{start}}$ and $\theta_{\text{fail}}$ is given Figure 5.4(b) for Dyneema® SK62. In efforts to ensure negligible pretension in the fiber at the start of the experiment, the slack-start procedure has been implemented, similar to that found in typical yarn tension tests [112]. The cross-head velocity used for all experiments is 1.2 mm/s, resulting in a nominal strain rate of 0.005/s. Finally, it is important to note that the angle $\theta_{\text{start}}$ is varied to ensure that geometric conditions similar to that found in transverse impact are enforced. In total, with variations in fiber type (five fiber types), indenter geometry (three indenters), and failure angle (five angles), roughly 700 experiments are performed, as each loading case is repeated ten times.

Single fibers with a gauge length of 557 mm are also pulled in longitudinal tension so as to measure a baseline failure strain without the induced stress concentration produced by the indenter. Single fibers are pulled in tension using the bollard grips in a similar fashion to that used in yarn tension tests [112]. Use of the bollard grips is employed so as to ensure a similar boundary condition to that used in the transverse loading environment. Thus, similar to the transverse loading experiments, carbon tape is affixed to the compressive platens so as to negate any grip slip. The
slack-start procedure is used in this testing sequence in efforts to minimize any fiber pretension. Displacement and load signals are gathered similar to the transverse loading experiments.

Finally, the fracture surfaces from both the Kevlar® KM2 and Dyneema® SK76 fibers are imaged via SEM (FEI-Nova) in efforts to gain insight into the failure mechanisms presented from both the aramid and UHMWPE fiber classes, respectively. The working distance and accelerating voltage for all samples ranges between $\sim5-9$ mm and 1-10 kV, respectively. All imaged samples are coated with a thin layer of AuPd in a Hummer 6.2 sputter coater in efforts to negate the presence of charging.

5.4 Results and Discussion

5.4.1 Transverse Loading - The effect of angle and indenter geometry on failure strain

Kevlar® KM2, Spectra® 130d, Dyneema® SK62, Dyneema® SK76, and Zylon® 555 are loaded in a transverse deflection environment as previously described and as shown in Figure 5.1 using three different indenter types, namely: FSP, razor blade, and round. Report of fiber longitudinal failure strain resulting from the transverse loading experiments using all three indenters at varying failure angles, $\theta_{\text{fail}}$, can be found in Figure 5.5. The traditional theory wherein failure strain is assumed to be minimally affected by loading angle is also overlaid on the plot as a dash-dot line, with values being identical to those shown in Table 5.1. This failure strain value is determined by filament longitudinal tension tests with a gauge length similar to that used for the transverse loading scheme. Use of axial failure strain rather than axial failure stress has been chosen so as to follow the traditional input failure criterion described by the classical Smith equations, which are shown below in Equations 5.2-5.3 [20].

For all fiber types studied, the trend in failure strain with increasing failure angle is consistent with the three different indenter geometries. The round indenter shows
Figure 5.4. (a) Representative indenter vertical displacement and vertical force exhibited onto the indenter as a function of time during testing. The experiment shown consists of a Dyneema® SK62 fiber loaded to a $\theta_{\text{fail}}$ value of 51°, using the FSP indenter. (b) Range in angle during transverse loading, from starting angle ($\theta_{\text{start}}$) to failure angle ($\theta_{\text{fail}}$), plotted against failure angle ($\theta_{\text{fail}}$). Data shown is from FSP loading of Dyneema® SK62 fibers.
negligible degradation to the failure strain as compared to the fiber longitudinal tensile failure strain, which is listed in Table 5.1, due to the minimal stress concentration presented on the fiber-indenter contact site. The razor blade indenter shows a direct reduction in failure strain as compared to the fiber longitudinal tensile failure strain. Furthermore, for fibers loaded with the razor blade indenter, there appears to be a negligible effect of failure angle on the resulting failure strain, though there is a slight correlation for the Kevlar® KM2 fiber. This sharp drop in failure strain can be attributed to an extreme stress concentration present at the indenter fiber interface, which may have reached a critical value, as there is minimal correlation with failure angle for all fiber types. It is important to mention that the radius of curvature of the blade is extremely fine, measuring roughly 2 μm, which can be seen in Figure 5.3(b).

Most intriguing to this study, the FSP exhibits a definite reduction in failure strain with increasing levels of $\theta_{fail}$, demonstrating that there is a clear stress concentration developed around such a geometry, which is directly affected by the test angle. Additionally, the failure strain as a function of $\theta_{fail}$ of the filaments loaded by the FSP indenter appear to be bracketed above and below by the resulting round indenter and razor blade indenter results, respectively. On the appropriate plots, cut data is overlaid from Mayo and Wetzel [129] and Shockey et al. [126], who analyzed the effect of a razor blade cutting through fiber and yarn, respectively. All data from Mayo and Wetzel that are marked in Figure 5.5 are derived from strength values [129], wherein failure strength data is converted to failure strain assuming the fiber exhibits a linear elastic loading response. Thus, failure strain is determined as $\varepsilon = \frac{\sigma}{E}$, wherein the $E$ utilized is that given in Table 5.1. Data from Shockey is taken directly from Shockey et al. [126]. As can be seen from the plots in Figure 5.5, the Mayo and Wetzel results line up quite well for Dyneema® SK76 while for both Kevlar® and Zylon®, the current data set lies above that found in Mayo and Wetzel, possibly due to a difference in fiber type (the current data set uses Kevlar® KM2 rather than Kevlar® 129 and the type of Zylon® used is not listed in the Mayo and Wetzel article) or a difference in blade sharpness both at the onset of testing and subsequent loading. The blades used
Figure 5.5. Experimental results of fiber longitudinal failure strain as a function of failure angle for three different indenter types for (a) Kevlar® KM2, (b) Spectra® 130d, (c) Dyneema® SK62, (d) Dyneema® SK76, and (e) Zylon® 555. Round, FSP, and razor blade indenters are implemented, represented by solid grey, solid black, and grey dotted lines, respectively.
in the current data set have a curvature radius of $\sim 2.3 \ \mu m$, while the blade shown by Mayo and Wetzel was measured at 1.5 $\mu m$. Furthermore, it has been reported that Zylon® yarn was able to deform a hardened steel utility blade [128], so it is possible that the blade used in the current setup may be softer than that utilized by Mayo and Wetzel, thereby being more easily blunted. It is also important to note that for an unknown reason, loading Zylon® filaments with razor blades is extremely difficult to perform as filaments fail during the sample loading process, thus only experiments at an angle of $10^\circ$ are shown.

Shin et al. assessed the effect of cutting through yarns composed of Zylon®, Spectra®, and Kevlar® and determined that Zylon® exhibited a much higher energy to break than both Kevlar® and Spectra® [128], which differs from Mayo and Wetzel, who found that Zylon®, Kevlar®, and Dyneema® all possessed a reasonably similar degradation in failure strength [129]. Mayo and Wetzel attribute the differences in the two studies to three factors [129]: (a) Zylon® has a higher elastic energy to break, thereby inherently increasing the level of absorbed strain energy for Zylon® fibers, (b) in yarn testing, the Zylon® force curve drops off slowly after peak loading while Kevlar yarns fail rather quickly post peak load, thereby increasing the level of energy dissipation for Zylon®, with such a delayed post-peak drop-off possibly arising due to the blade tilt, and (c) inter-fiber friction could keep the Zylon® yarn from flattening out during cutting, thereby keeping fibers away from the sharp cut surface. In light of such discrepancy, the effect of blade tilting needs further analysis in efforts to better analyze cutting events that may occur to a soft armor system during use, such as slashing caused by a knife blade.

It is of importance to note that the angle created during an actual transverse impact event is due to the impacting velocity and material properties of the fiber, not exclusively the indenter head [20]. A schematic of the transverse impact event can be seen in Figure 5.6, and is described by Equations 5.2-5.6.

Upon contact of the projectile with the yarn, a longitudinal stress wave is developed and moves at the material sound speed, described by $c$ in Equation 5.2
Figure 5.6. Schematic of transverse impact of an FSP into a high-performance yarn/fiber.
\[ c = \sqrt{\frac{E}{\rho}} \]  
(5.2)

wherein \( E \) and \( \rho \) are the longitudinal elastic modulus and the fiber density, respectively. Material behind the wake of the longitudinal wave is set in motion towards to impact sight, moving with a particle velocity described by \( W \) in Equation 5.3,

\[ W = C\epsilon \]  
(5.3)

where represents the longitudinal strain developed in the filament. At the time of impact a transverse wave front is also emanated from the impact site moving at a velocity \( U \) being described by Equation 5.4.

\[ U = \sqrt{\frac{E}{\rho}} \left( \frac{\epsilon}{1+\epsilon} \right)^{1/2} \]  
(5.4)

Material in the wake of the transverse wave front is set in motion in the direction of the impacting projectile, moving with a particle velocity \( V \) being described in Equation 5.5, which is identical to the projectile impact velocity.

\[ V = ((1 + \epsilon)^2U^2 - ((1 + \epsilon)U - W)^2)^{1/2} \]  
(5.5)

If Equations 5.2 - 5.5 are solved simultaneously, a velocity causing immediate failure (instantaneous rupture velocity) can be determined if one knows the fiber breaking strain. Finally, the angle developed during impact is described by \( \theta \) in Equation 5.6.

\[ \theta = \tan^{-1} \left( \frac{V}{U(1+\epsilon) - W} \right) \]  
(5.6)

It is postulated that such angles undergone during transverse impact would produce sharp kinks or local stress concentrations even with a rounded indenter at the leading edge of the transverse wave front. This is evidenced by the instantaneous lo-
cal failure of UHMWPE yarns when impacted by high velocity 5.5 mm steel spheres, wherein it was found that the transverse impact event produced fibers bearing localized failure surfaces that had been sheared through their cross-sectional thickness [50], which is similar to the razor blade and high $\theta_{\text{fail}}$ results from the FSP indenter shown in Section 3.2. Such a kink angle is not developed for rounded indenters during the current quasi-static testing setup, as the fiber will always leave the rounded indenter head tangent to the indenter surface curvature. Thus, the rounded indenter experiments performed here more appropriately mimic pure longitudinal tensile testing, and the resulting similarity in failure strain demonstrates the lack of stress concentration developed at both the indenter head and the bollard grip contact site. Thus it is postulated that the stress concentration developed during a transverse impact event can be reasonably approximated using the current results listed for the FSP indenter.

In order to visualize the difference in failure strain degradation with increasing failure angles between the five fiber types, FSP data from Figure 5.5 are re-plotted in Figure 5.7. From this set of data, it can be surmised that all fiber types tested show a demonstrative effect due to the angle of failure. It is also important to note that error bars have been omitted from the presented curves so as to make the figure more clear, but said error bars are presented in Figure 5.5, and the reader is directed to said figures if error bars are of interest.

As can be seen, while all fiber types exhibit a demonstrative coupling to the failure angle, the UHMWPE and Zylon® fibers have a higher failure strain for the entire range of failure angles tested when compared to Kevlar® KM2. The Kevlar® KM2 fiber also presents the highest degree of degradation due to the change in angle when loaded by the FSP. It is also important to note the unexpected low value of failure strain exhibited by Kevlar® and Zylon® when pulled in pure tension ($0^\circ$ tension tests). The zero degree tests were conducted a second time in efforts to ensure that the data is not due to testing error conditions. Furthermore, the fibers fail in the gauge section for all listed pure tension tests. When using the FSP indenter, it is found that failure always occurs at either edge of the indenter, rendering the length
of the fiber affected by the stress concentration quite localized, as failure is not seen to occur randomly throughout the gauge length, but at the indenter corner. Most importantly, it is imperative to reiterate that the FSP loading geometry drastically affects the resulting longitudinal failure strain of all fibers tested, especially at higher loading angles.

Reasoning for such a reduction in failure strain could be due to a number of scenarios, such as an increased shear loading between fibrils, a detrimental compressive stress felt under the filament neutral axis which ultimately causes kink bands, an additional tensile stress at the top surface of the fiber due to bending, or even abrasive friction along the length of the fiber. Regardless, the presence of such degradation is clear, which shows the effect of loading a fiber in a multi-axial environment. It is imperative to state that such failure in this quasi-static regime is clearly a function of the indenter shape, while such geometry is of no consequence during transverse impact. Carr has shown that both aramid (Kevlar® 129 and KM2) and UHMWPE (Dyneema® SK66) fibers exhibit a greatly reduced instantaneous rupture velocity when impacted by 5.5 mm steel spheres [50], being quite similar to results by Chocron et al. [53] who used an FSP projectile during transverse impact experiments. Such an absence of geometrical effects yields the current authors to postulate that failure could also be caused at the kink created by the transverse wave front, though no immediate effort has been made to verify such a suggestion. Regardless, clearly there is an effect of both loading angle and indenter geometry during quasi-static transverse deflection and further analysis is most definitely required to determine if such an effect is also present during actual transverse impact experiments. Additional reasoning of such a failure process is provided in Hudspeth et al. [139].
Figure 5.7. Shown in this plot is the comparison between the failure strain as a function of angle using an FSP indenter for Kevlar® KM2, Spectra® 130d, Dyneema® SK62, Dyneema® SK76, and Zylon® 555 (data identical to that found in Figure 5.5). Data for 0° has been taken from longitudinal tension tests and is identical to the failure strain listed in Table 5.1.
5.4.2 Transverse Loading - The effect of angle and indenter geometry on filament failure surfaces

In efforts to gain insight into the effect of the changing loading geometry, fracture surfaces from a representative aramid (Kevlar® KM2) and UHMWPE (Dyneema® SK76) are imaged via high-resolution SEM (FEI-Nova). Rupture surfaces from all three indenters are analyzed for 10°, 20°, 30°, 40°, and 50°, with representative fracture morphologies being presented in Figures 5.8 and 5.9 for Kevlar® KM2 and Dyneema® SK76, respectively.

For the Kevlar KM2 fiber, the failure surfaces match very well with the failure strain data presented in Figure 5.5(a). The rounded indenter, which promoted a reasonably constant failure strain, produces similar fracture pattern for all tested angles, being defined by fibrillation, which is the typical failure surface found in aramids when pulled in pure axial tension [38]. Similarly, the razor blade indenter, which also caused a relatively constant failure strain value for all of the tested angles, produced very similar fracture patterns at all $\theta_{\text{fail}}$, being defined by a localized failure, seeming as though the blade shears through the fiber thickness, along with a slight degree of axial splitting due to the extremely poor transverse bonding between fibrils. Such local failure due to transverse cutting has been previously documented by several authors [126,128,129], albeit the current failure surfaces do not exhibit the locally splayed out pattern that is sometimes presented during low angle transverse cutting. It is also important to note that at lower $\theta_{\text{fail}}$ geometries, it looks as though the majority of the fiber is cut through the cross-section thickness, with the final remaining portion looking like a ribbon which fails either by being stripped off along the surface until a critical defect is reached or more simply via tensile fibrillation failure. Reasoning for such a failure process may be due to local yielding/dulling of the razor blade, which could push fibrils to the point of separating from the main fiber body until ultimate failure. Most interestingly, the FSP indenter, which as aforementioned caused a reduced failure strain with increasing $\theta_{\text{fail}}$, likewise promotes a changed
fracture pattern with increasing $\theta_{fail}$. At low $\theta_{fail}$ values, the fracture morphologies exhibit definite fibrillation, similar to that produced from the rounded indenter. At high $\theta_{fail}$ values, the post-mortem rupture surfaces possess a much more local failure, looking quite similar to the samples loaded by the razor blade indenter. Furthermore, it is important to note that even with the measured radius of curvature of the FSP being roughly twice the diameter of the Kevlar KM2 fiber, there must exist a definite local stress concentration around the contact boundary, as the fiber clearly transitions from a long-range fibrillated failure surface to a much more local rupture that looks like it is sheared through the filament thickness.

Similar to the Kevlar KM2 failure surfaces, the Dyneema® SK76 results follow quite well with the failure strain results presented in Figure 5.5(d), albeit the Dyneema® fracture surfaces do not present as stark of a transition from fibrillation to local shearing, though the variation is still present. As can be seen, loading by the rounded indenter, which causes a reasonably unchanged fiber failure strain, produces a fibrillated microstructure with a reasonable level of axial splitting similar to that seen in pure longitudinal tension tests [38]. It is important to note that at higher $\theta_{fail}$, the fracture patterns produced by the rounded indenter do seem to possess fibrils which have deformed into slightly bulbous shapes, which may possibly be due to snap-back of the fiber post failure. The razor blade indenter, which also produces a relatively unchanged failure strain, promotes a similar rupture morphology for all tested failure angles, being defined by a local through thickness shearing failure. Similar to the Kevlar® failure, at low $\theta_{fail}$ the razor blade seems to cut through the majority of the fiber via local shearing. The final failure looks to promote a longer range artifact being caused by either the aforementioned ribboning or because of fibrillation. Regardless, this final failure effect is believed to be caused by blade dulling. Again, as with the Kevlar® KM2 micrographs, the effect of the FSP on the Dyneema® SK76 failure surfaces shows an evolution from fibrillation/axial-splitting at low failure angles, being similar to failure surfaces caused by the rounded indenter, to short-range cutting/shearing, being quite similar to the razor blade micrographs. Clearly, the
effect of the FSP corner instigates a local stress concentration into the fiber as it is pulled in tension, demonstrated by the rupture morphology evolution from longer range fibrillation to short range shearing at increasing values of $\theta_{\text{fail}}$. It is also important to note that such short range failure in UHMWPE fibers has also been recorded by filaments loaded via transverse impact up to velocities causing instantaneous fiber rupture [50]. Interestingly, the projectile used was a steel sphere. From the current rounded indenter micrographs, it is concluded that such a shearing failure mode is not caused by a rounded indenter. Rather, this mode of shear failure is caused by a local sharp angle produced by a small radius of curvature corner, demonstrating that such shearing effects may not be caused by the projectile geometry, but rather by the geometry produced from the transverse wave propagation during transverse impact.

5.5 Stress Analysis

In efforts to identify possible failure zones demonstrated by the fiber about the projectile head contact site, a quasi-static FE analysis in Abaqus Implicit has been designed specifically around the previously mentioned setup shown in Figure 5.1 with an emphasis placed on recreating similar long-range boundary conditions as to those developed in experimental results. Specifically, Figure 5.10 demonstrates a schematic of the FE analysis of the $35^\circ$ loading condition using an FSP indenter. It is important to note that effort has been placed on attaining boundary conditions and loading conditions for the FE analysis that resemble the experimental conditions as closely as possible.

A 2-D planar model was selected to reduce run times, employing roughly 500,000 planar triangular elements. A transversely isotropic material model has been applied to the fiber, containing the material properties shown in Table 5.2. The indenter was modeled as a rigid solid, possessing radius of curvatures determined by the polished cross-sections shown in Figure 5.7. Contact conditions allow for projectile-fiber interface sliding with an applied friction coefficient of 0.2 [143]. It is important to
Figure 5.8. SEM micrographs taken of the various rupture surfaces presented by the KM2 fiber when loaded by the three different indenters and failing at the described failure angles. The failure surfaces match very well with the failure strain data presented in Figure 5.5(a).
Figure 5.9. SEM micrographs of the rupture surfaces developed by the Dyneema SK76 fiber when loaded with the three different indenters at the various failure angles. Similar to the KM2 failure surfaces, the SK76 results follow extremely well with the failure strain results presented in Figure 5.5(d).
Figure 5.10. Schematic of FEA model used to analyze local stress states developed around the indenter head.
Table 5.2. Material properties of Kevlar KM2® used in the transversely isotropic material model employed in both quasi-static and dynamic regime.

<table>
<thead>
<tr>
<th>Material</th>
<th>$E_{11}$ (GPa)</th>
<th>$E_{22}$ (GPa)</th>
<th>$G_{12}$ (GPa)</th>
<th>$\nu_{12}$</th>
<th>$\nu_{23}$</th>
</tr>
</thead>
<tbody>
<tr>
<td>Kevlar® KM2</td>
<td>84.62</td>
<td>1.34</td>
<td>2.4</td>
<td>0.6</td>
<td>0.24</td>
</tr>
</tbody>
</table>

Figure 5.11. Resulting long-range modeled stress profiles (long-range being defined as a location being far away from the stress concentration effects demonstrated around the projectile-filament contact site)

Note that the long-range tensile stresses felt by the modeled single filament (shown in Figure 5.11) have been forced to mimic those which are seen at failure from experimental results, which is shown in Figure 5.5a. Due to the large deflections inherent in the model, run times were roughly 30 minutes (wall clock), using an 8-core CPU and 32 Gb of memory. Upon adequate achievement of said long-range stress state, specific attention has been placed on tracking the longitudinal, transverse, and shear stresses developed in the yarn material directly around the projectile contact site, and resulting color intensity plots of stress-state profiles can be seen in Figure 5.12.

In Figure 5.12 can be seen stress field plots generated for single fiber quasi-static loading. No color intensity magnitude bar has been given immediately in Figure 5.12, as the color maps have been included solely to provide an understanding of stress state distributions. Negative stresses are depicted in blue and positive stresses
in red. Additionally, three hypothesized zones of potential failure have been selected, as they depict max tensile, compressive, or shear stress states. Max/min stress values present within said zones are measured from each profile shown in Figure 5.12, and have been subsequently plotted for each of the three indenter geometries, being listed in Figure 5.13.

In Figure 5.13, max/min stress intensities have been plotted for the local region around the indenter-filament contact site. Figures 5.13(a), 5.13(b), and 5.13(c) depict max/min S11, S22, and S12 values, respectively. The 1-, 2- coordinate system has been locally defined on the element level to be along the longitudinal and transverse fiber directions, respectively. As shown in Figure 4(a-c) the round projectile head has very little effect on the local stress-state, with all elements showing a nearly identical stress-state of a positive (tension) S11 value and negligible S22 and S12 stresses, which is to be expected, due to the large radius of curvature demonstrated from the round indenter head. In contrast, the razor blade and FSP indenters demonstrate similar S11, S22, and S12 stress profile trends. S11 tensile stresses are felt on the top of the fiber above the projectile contact site in zone 1, and show little variation due to changes in loading angle. In contrast, S11 compressive stresses are largely affected by changes in fiber loading angle, with increasing compressive stresses felt at increasing loading angle. Maximum longitudinal compressive stresses are shown to reside at the bottom of the fiber located just above projectile contact site in zone 2. Similar S22 stress profile histories are also demonstrated for both the FSP and razor blade indenters, although much more drastic for the razor blade. Slight increases in transverse tensile stresses are felt in zone 3, roughly 45° angle from the projectile-filament contact site, with increasing transverse tensile stresses being demonstrated at increasing fiber loading angles. Large transverse compressive stresses are developed directly in the vicinity of the projectile-filament contact site in zone 2, with the FSP stress state staying reasonably constant when subjected to changes in fiber loading angle, while the razor blade indenter shows an increase in transverse compressive stresses with increasing fiber loading angle. Finally, both FSP and razor
Figure 5.12. Color intensity plots of (a) S11, (b) S22, and (c) S12 stress profiles around the edge of an FSP projectile-indenter contact site loaded to 35°. 1-, 2- directions are locally oriented along the fiber longitudinal and transverse directions, respectively. Legend units are in N/μm².
Figure 5.13. (a) S11, (b) S22, and (c) S12 stress states developed in single filament, quasi-static transverse loading for round, FSP, and razor blade indenters.

Blade indenters yield a definite variation in the S12 shear stress due to increasing fiber loading angle, with max shear stress values being demonstrated in zone 3, roughly $45^\circ$ from the indenter-filament contact site.
5.5.1 Effects of Friction

In efforts to assess the effect of the applied indenter contact friction law on the resulting internal stress fields created in the fiber, the indenter-filament friction coefficient has been altered by roughly an order of magnitude both above and below the previously mentioned value of 0.2. Friction coefficients of 0.02, 0.2, and 1.0 were analyzed and resulting stress profiles from each condition are shown in Figure 5.14 when using the FSP indenter profile at a variety of loading angles.

Figure 5.14(a) demonstrates the long range stress developed in the fiber at distances well away from the contact site. At such distances, stresses in these far region zones should be independent of the contact law present in the local zone around the indenter contact site, which is indeed verified as shown in Figure 5.14(a). Figures 5.14(b), 5.14(c) and 5.14(d) demonstrate the local variation in $S_{11}$, $S_{22}$, and $S_{12}$ stress states immediately in the vicinity of the FSP. $S_{11}$, which is locally oriented in the longitudinal direction of the fiber, demonstrates identical max longitudinal tensile stresses independent of friction coefficient, which are present in zone 1 shown in Figure 5.12(a). Max longitudinal compressive stresses, which are found in zone 2 (Figure 5.12(a)) show a slight increase in longitudinal compressive stress magnitude with increasing friction coefficient, but the variation is quite small. $S_{22}$, which is locally oriented in the filament’s transverse direction, shows no perceptive variation for both max tensile and compressive values, being found in zones 3 and zones 2 defined in Figure 5.12(b). Finally, values for $S_{12}$ are shown to also be affected in a negligible fashion by the analyzed variation in friction coefficient. The largest magnitude shear stresses are found to reside in zone 3, as shown in Figure 5.12(c). Thus, due to the very small effect played by friction coefficient in the current analysis set, the implemented friction coefficient value of 0.2 is deemed to be reasonable within the current data analysis set.
Figure 5.14. The effect of friction on (a) Long range S11, (b) local S11, (c) local S22, and (d) local S12 stress states developed in single filament, subjected to various loading angles.
Table 5.3. Mesh sensitivity study of both FSP (35°) and razor blade (21°) simulations.

<table>
<thead>
<tr>
<th>Indenter</th>
<th>Angle (deg.)</th>
<th>Number of Elements</th>
<th>S11 Max/Min (GPa)</th>
<th>S22 Max/Min (GPa)</th>
<th>S12 Max/Min (GPa)</th>
<th>Wall Clock (min)</th>
</tr>
</thead>
<tbody>
<tr>
<td>FSP (Coarse)</td>
<td>35</td>
<td>55011</td>
<td>5.08/-6.223</td>
<td>0.222/-0.654</td>
<td>0.0385/-0.95</td>
<td>31</td>
</tr>
<tr>
<td>FSP (Fine)</td>
<td>35</td>
<td>210021</td>
<td>5.06/-6.31</td>
<td>0.236e/-0.657</td>
<td>0.0467/-1.04</td>
<td>97</td>
</tr>
</tbody>
</table>

5.5.2 Mesh Convergence

A mesh convergence study was also undertaken to verify appropriate mesh fineness in efforts to achieve sufficient output accuracy while reducing wall clock times of computation runs. It must be noted that much more effort was placed on verifying accurate results than minimizing model run times as it was possible to run several sequential model variations over long periods of computer downtime. As such, the following results are focused mesh fineness verification, with a byproduct of demonstrating wall-clock time reductions. The FSP, 35° condition was selected for the first mesh analysis and resulting max/min stress states and corresponding wall clock times are compared against the number of elements present in the model, with results being shown in Table 5.3.

5.5.3 Failure Zones

At this point, it is important to discuss the possible failure zones of the filament material, which are defined as zones 1, 2, and 3, as shown in Figure 5.12. Although the highest equivalent stress state is felt within zone 2, it must be noted that this stress state is entirely compressive in nature. It is quite possible that kinking could develop in the filament within zone 2 due to longitudinal compressive stresses. Although typically regarded as unassociated with reduction in subsequent longitudinal tensile strength [37,38,132,144], kinking sometimes reported to reduce a filaments longitudinal tensile failure strain [143]. That said, it is important to remember that the longitudinal compressive stress developed in the filament is due to bending, and
as such this stress state is not subsequently put in tension, thereby rendering local
damage effects within the kink zone to more than likely be irrelevant. Furthermore,
it is important to remember the high transverse compressive stress felt within zone
2, likely attenuating kink development. Due to such factors, it is suggested that it is
that failure initiating from zone 2 is doubtful. Rather, it is suggested that failure more
than likely occurs in either zones 1 or 3. Zone 1 experiences the max tensile stress
felt within the fiber, due to high localized bending that occurs around the projectile
corner, and it is completely feasible that failure initiates within the region of max
tensile stress state present at zone 1. It is also quite possible that filament failure
occurs within zone 3, as the fiber experiences a combination of transverse tension
and local shearing within said zone. Such a combination of shearing and tension may
indeed promote local rupture of the filament as previous works have shown that fiber
rupture actually occurs by intermolecular shear, rather than chain scissioning [29].
As such, the tensile stress state within the fiber could be greatly compounded by the
high shear-stress presence, thereby initiating failure at zone 2. That said, within the
current scope of the modeling analysis, it is difficult to deduce the actual location
of failure, and future effort should be made to directly understand the material fail-
ure zone. It also must be pointed out that this material model does not currently
contain plastic deformation, which could decidedly change both the bending stresses
developed in zone 1 and the resulting shear stresses seen in zone 3, due to flattening
of the fiber around the projectile head, resembling a race track cross-section [145].
Regardless, it is still postulated that even in the presence of filament plastic defor-
mation, zone 1 will demonstrate high longitudinal tensile stresses and zone 2 will feel
a combination of local shearing and longitudinal tension.

5.6 Conclusions

Single high-performance fibers are loaded in a transverse deflection environment in
efforts to gain insight into the effect of large deflection angles on the resulting filament
longitudinal strain at failure. Further analysis on the effect of indenter geometry is explored by using a 0.30 caliber rounded head, a 0.30 caliber FSP, and razor blades. The fiber types explored in this study are Kevlar\textsuperscript{\textregistered} KM2, Spectra\textsuperscript{\textregistered} 130d, Dyneema\textsuperscript{\textregistered} SK62, Dyneema\textsuperscript{\textregistered} SK76, and Zylon\textsuperscript{\textregistered} 555. For all fiber types experimentally studied, when using the rounded indenter, there is a negligible reduction in longitudinal fiber strain with increasing angle, with the strain value being quite similar to the respective longitudinal tensile failure strain resulting from tension experiments implementing a similar gauge length of 0.56 m. Fibers loaded by the razor blade show a demonstrative decrease in strain to failure as compared to the rounded indenter, but there is little to no effect of loading angle on the resulting failure strain. Most interestingly, the fibers loaded by the FSP show a large effect from the loading angle on the resulting longitudinal fiber strain at failure. While the low angle experiments yield results quite close to that from the rounded indenter, with increasing loading angle, there is a stark drop in the strain-to-failure. At the largest angles tested, the failure strain is quite close to the results from the razor blade indenter. In addition to tracking the strain-to-failure for the different loading conditions, post-mortem fracture surfaces from Kevlar\textsuperscript{\textregistered} KM2 and Dyneema\textsuperscript{\textregistered} SK76 are imaged in order to gain further insight into the fracture mode exhibited for the various experiments. Rupture morphology evolution follows extremely well with the strain-to-failure results. The rounded indenter promotes rupture surfaces exhibiting long-range fibrillation similar to that found in pure longitudinal tension tests, while the razor blade is found to enforce short range failure caused by through-thickness shearing. As with the strain-to-failure results, the FSP indenter promotes a variation in failure mode with increasing deflection angles; fibrillated fracture surfaces are found at low deflection angles, while local shearing is demonstrated at high deflection angles. From both the strain-to-failure results and rupture morphology trends, it is clear that there exists a demonstrative effect of loading conditions on the failure strain and failure mechanism of high-performance fibers. Clearly the assumption of pure longitudinal tension properties being the sole indicator of ballistic performance of high-performance fibers
must be reconsidered. To the authors’ knowledge, such an assumption is used in the all-analytical and constitutive models for fiber, yarn, and fabric, especially for soft body armor systems.
6. Effect of Projectile Nose Geometry on the Critical Velocity and Failure of Yarn Subjected to Transverse Impact

Adapted from:

6.1 Abstract

Three different types of yarn have been subjected to transverse impact experiments in efforts to gain an understanding of local yarn failure and to provide input parameters for future transverse yarn impact simulations. Dupont™ Kevlar® KM2, DSM Dyneema® SK76, and AuTx® from JSC Kamenskvolokno were selected as representative materials as the former two are commonly implemented into bullet resistant panels and the latter is a promising material for future impact resistant fabrics. In order to assess the effect of projectile nose shape on the critical rupture velocity range for each yarn type, three missile geometries have been implemented, namely 0.30 caliber rounded head, 0.30 caliber chisel nosed fragment simulation projectile (FSP), and high-carbon steel razor blade. As opposed to one single velocity wherein yarn behavior transitions from transverse wave development to immediate local failure, a range is defined wherein progressive filament failure is detected with increasing impact velocities. Such ranges are determined for all yarn types using the three projectile geometries yielding critical velocity transition regions of increasing value when impacting via razor blade, FSP, and round projectile heads, accordingly. Additionally, post-mortem fracture surfaces recovered from impact experiments have been imaged so as to elucidate the mechanism of failure throughout the range of ve-
locities tested for each projectile type and yarn material and said fracture surfaces correlate well with impact velocity and projectile nose geometry.

6.2 Introduction

High-performance polymeric fiber is used in a plethora of applications due to its unmatched stiffness-to-weight and strength-to-weight ratios. While industrial applications and sporting equipment such as fishing line, climbing ropes, water vessels, and tires are areas of interest, the direction of the current work is entirely focused on ballistic functions such as body armor and turbine fragment containment systems. Due to the inherent life-saving nature of these applications, it is clearly of great concern for researchers and designers to understand the mechanism by which said materials halt incoming projectiles. Below the ballistic limit of a fabric system, energy dissipation mechanisms chiefly exhibited by principal yarns have been shown to play the major role in halting the incoming projectile, being defined by longitudinal strain energy, longitudinal kinetic energy, and transverse kinetic energy [6]. These energy mechanisms have been shown to be governed by effects such as, but not limited to, fiber-fiber friction, yarn-yarn interaction (sliding, trellising, pull-out), weave structure, environmental aging, projectile nose geometry, and projectile strike angle [7, 8]. In contrast, above the ballistic limit of a fabric system, various authors have shown that the effect from many of these mechanisms are attenuated, and failure becomes localized or ‘inelastic’ [7, 9–11]. Such an immediate local failure process inherently prevents energy from moving away from the impact site, as both longitudinal and transverse stress wave propagation is undeveloped due to early or premature impact site failure, yielding the consequence of an almost negligible projectile deceleration [12, 13].

This stark transition in energy dissipation from full fabric panel involvement at relatively low impacting velocities to rapid local failure exhibited at relatively high impact velocities is not only governed by projectile velocity, but is also coupled to projectile nose geometry, which directly affects both the mode of penetration and
the constituent yarn failure. Montgomery et al. [14] and Tan et al. [15] showed that increasing nose sharpness of impacting projectiles into aramid fabrics allowed for more pronounced fabric windowing/nosing, thereby decreasing the ballistic limit of a fabric system when impacted by increasingly sharp-nosed projectiles. Additionally, Tan et al. [15] demonstrated that flat nosed projectiles were more adept at cutting through the constituent yarns while spherical head impact resulted in longitudinal tensile failure of the yarns beneath the fabric footprint. The sharpness of a projectile nose geometry also controls the ballistic limit of a fabric system, indeed, Abbott demonstrated that both Nylon 6/6 composites and an undisclosed dry aramid fabric revealed $V_{50}$ values coupled to the chamfer angle of both conical nosed projectiles and FSPs [16]. It was alluded that the reduction in armor $V_{50}$ due to sharpened conical projectiles resulted from increased windowing through the fabric weave, while the decrease in fabric $V_{50}$ due to increasing FSP chamfer angle was described as ‘expected’, and reference to Prosser’s [9] explanation of local shearing was mentioned. The cutting effect by sharp-cornered projectiles has also been described by [7] to be driven by local shearing of the constituent yarns, wherein he notes a clear difference is seen when impacting a variety of fabric types/thicknesses with chisel-nosed FSPs and right circular cylinders (RCCs). Interesting to note is the effect of nose geometry on extremely overmatched fabric systems; Gibbon et al. (2014) impacted varying thicknesses of Twaron fabric packs with a variety of different projectile nose geometries at a constant overmatched velocity, and although stated otherwise in report, found nearly similar residual velocities for all projectile types. Additionally, both the number of broken yarns found on the impact surface ply and the full pack energy absorption were seemingly unchanged for all pack thicknesses (1-5 layers). Results from the immediately aforementioned studies [7, 9, 14–17] thus suggest that the response of the fabric system is governed by both structural artifacts and constituent yarn mechanical behavior.

In this light, it is of interest to separate the fabric’s structural and material behavior in efforts to gain any possible understanding of fabric failure in a fashion as
simple as feasibly possible. It is thus proposed in the immediate work to clear away
the structurally governed energy dissipation mechanisms which can occur both below
and near the ballistic limit of a fabric system and rather, focus on the local failure
of the constituent yarns. As transverse impact of single yarns promotes similar wave
fronts to those developed during the transverse impact event in full fabric, and be-
cause single yarns have also been shown to exhibit the ‘inelastic’ failure mode above
some critical impact velocity \[50,52,53\], it is suggested to look at the more tractable
single yarn impact event with an emphasis placed projectile nose geometry, which to
the authors’ knowledge, has not been systematically studied in existing literature.

First and foremost, a brief description of the geometry exhibited during transverse
impact into yarn is required. Although rightfully attributed in the United States to
a series of Smith et al. [20] studies, description of yarn deformation during transverse
impact has been historically developed by a variety of authors [116–118,120]. A
schematic of the impact event can be seen in Figure 6.1(a), and is described by
Equations 6.1-6.5. Such an experiment is performed by shooting a projectile with
velocity \(V\) along a shot line perpendicular to the yarn longitudinal direction. Upon
contact of the projectile with the suspended yarn, a longitudinal wave is developed,
moving outwards away from the impact site with the wave velocity \(c\), identical to the
speed of sound in the yarn. This wave velocity is described by Equation 6.1 where
\(E\) and \(\rho\), represent the yarn longitudinal elastic modulus in tension and yarn material’s
mass density, respectively.

\[
c = \sqrt{\frac{E}{\rho}}
\]  

(6.1)

Material behind the wake of the longitudinal wave is set in motion with a particle
velocity of \(W\) towards the impact site. The magnitude of this inward particle velocity
is defined in Equation 6.2, where \(\varepsilon\) represents the increase in longitudinal strain
developed in the yarn.
\[ W = C \epsilon \]  

(6.2)

At the time of impact an additional wave is also developed, namely the transverse wave, being geometrically described by a tent-like shape. The velocity of this transverse wave is described by Equation 6.3.

\[ U = \sqrt{\frac{E}{\rho}} \left( \frac{\epsilon}{1 + \epsilon} \right)^{1/2} \]  

(6.3)

Material that was initially moving inward towards the projectile with particle velocity \( W \) transitions to a velocity \( V \) in the direction of the projectile as it is passed by the transverse wave front, with \( V \) being defined by Equation 6.4.

\[ V = ((1 + \epsilon)^2 U^2 - (1 + \epsilon)U - W^2)^{1/2} \]  

(6.4)

The geometric angle \( \theta \) of the transverse wave front with a linear elastic yarn, which remains unchanged during the initial time of impact before reflections from the boundary, is defined by Equation 6.5.

\[ \theta = \tan^{-1} \left( \frac{V}{U(1 + \epsilon) - W} \right) \]  

(6.5)

Although a rather obscure experimental setup, the described impact geometry is well substantiated from a number of experiments. Smith et al. [57] shot high-strength Nylon yarn at velocities around 500 m/s, tracking transverse wave front development via high-speed imaging and found the resulting propagation to be quite similar to theory listed in Equations 6.1-6.5. Petterson looked at spatial distribution of strain along the yarn length during the impact event decisively showing that the strain distribution is unaffected by the passage of the transverse wave front and that the longitudinal wave speed is independent of projectile striking velocity [146]. In subsequent study, Petterson et al. [147] developed a means of ascertaining dynamic stress-strain curves of plastically deforming yarns using empirical knowledge of strain
vs. strain-velocity curves. Wilde et al. elucidated the energy loss of a projectile due to the retarding force acting on the projectile when impacting single yarns [58, 148]. When subjected to various levels of longitudinal pre-stress, Field and Sun measured transverse wave speeds from both Kevlar® and Spectra®, which agreed with theory quite remarkably [41]. More recently, Chocron et al. [53] also found measured transverse wave speeds to agree well with theory when impacting Kevlar®, KM2, Dyneema® SK65, and PBO yarn. Equations 6.1-6.5 were also verified by Song et al., who tracked both transverse wave speeds and resulting transverse wave angles when impacting single Kevlar KM2 yarns at velocities up to 324 m/s [149]. In addition, Figures 6.1(b) and 6.1(c) from the current work depict image sequences of Dyneema® SK76 and AuTx® yarn being impacted by a FSP and round projectile at velocities of 460 m/s and 640 m/s, respectively. Both image sequences have been successfully overlaid with the transverse wave front described within Equations 6.1-6.5.

Interestingly, as the incident striking velocity is increased for the yarn material, there exists a strike condition wherein the aforementioned waveforms do not develop as the yarn fails at the instant of impact. Examples of this peculiar critical velocity found throughout various literature are listed in Table 6.1. Due to the previously described geometric success, it is reasonable to assume that one could use the analytical relationships described by Equations 6.1-6.5 to determine the velocity at which a yarn would fail upon contact due to the longitudinal strain, $\varepsilon$, developed behind the longitudinal wave front. In reality, experiments demonstrate this critical velocity to be 40-50% below that which is predicted [50,52,54,121]. In efforts to reconcile such differences, three theories in literature have developed for modern fiber impact, namely multiple-point contact [52], bouncing [54], and multi-axial stress [139] and [143]. The first theory, multiple-point contact, is directed at spherical projectile impact and assumes the yarn is contacted at multiple locations along the surface of the missile. Summing up the strains developed from the multiple impact locations yields a cumulative strain state higher than that developed from a single point contact. Although
Figure 6.1. (a) Schematic of a transverse impact event into a single yarn. Such geometry is developed below the transverse critical velocity. (b) Impact of a 0.30-cal FSP into Dyneema® SK76 yarn at a velocity of 460 m/s, which is near the lower end of the critical velocity transition region (described below). (c) Impact of a round projectile into AuTx® yarn at a velocity of 640 m/s. Displayed images for (b) and (c) are spaced by 1 and 2 s, respectively (actual camera frame separation was 200 ns, but such resolution has been omitted from the current figure to provide visualization of greater deformation). White overlaying line segments in both (b) and (c) represent the transverse wave front as predicted by Equations 6.1-6.5. Note, frame orientation in (b) and (c) has been rotated 90° clockwise with respect to (a) and each frame size has been adjusted to contain the entire portion of yarn taken up by the transverse wave front.
mechanically sound, such a theory ultimately predicts a critical velocity of zero for a round projectile contact, as infinite contact locations exist around the yarn-projectile contact arc. Following such theory, Walker and Chocron [54] looked at the impact of a yarn by a FSP and point out that for a flat faced impact condition, the yarn should bounce off the face of the impactor, thus leading to two contact locations for the projectile-yarn interface. Longitudinal stress waves emanate from both contact locations, and with the meeting of the inward moving waves, an increased stress state in front of the projectile is developed. Also mechanically sound, such a theory does not account for impact with sharp or rounded indenters, which as will be shown shortly, both exhibit a decrease in the velocity that promotes immediate yarn rupture when compared to that predicted from Equations 6.1-6.5. In order to circumvent the effects of bouncing or wave mechanics, Hudspeth et al. [139] looked at the effect of loading single Kevlar® KM2 fibers in a geometry similar to that which would be developed at the experimental critical velocity found by Chocron et al. [53]. It was determined that the fiber exhibited a decrease in failure strain due to a local stress concentration developed at the corner of the indenter contact site, with said strain being quite close to the failure strain demonstrated in actual transverse impact experiments. Furthermore, numerical simulation of single-fiber transverse impact from Sockalingam [143] corroborates the notion of a multi-axial stress development in the fiber, being demonstrated by a high bending stress in the fiber near the projectile contact site. Indeed, as briefly noted by Wilde et al. in experimental report, preliminary experiments comparing sharp cornered projectiles to projectiles having 0.2-in ground radius demonstrate a reduction in both yarn breaking velocity and the resulting projectile energy loss (Wilde et al., 1970; Wilde, 1974). In this light, it is thus proposed that the shape of a projectile does indeed have an effect on the resulting transverse impact critical velocity of a yarn. It must also be noted that additional efforts have also tried to explain the variation demonstrated in transverse critical velocity, but focus was directed towards fibers that were highly rate-sensitive and inelastic [150,151].
Table 6.1. Critical velocity of various yarns. Note: Guided Projectile Transverse (GPT) Impactor mimics a notched saddle geometry. *Carr states that Kevlar K129, Kevlar KM2, Twaron CT yarns are used for experiment, although notice is not explicitly given for which aramid is used at each data point. **Carr has used both 440 dtex and 880 dtex yarns for experiment, with no notification given for the origin of each data point.

<table>
<thead>
<tr>
<th>Author</th>
<th>Material</th>
<th>Projectile Geometry</th>
<th>Velocity Range (m/s)</th>
</tr>
</thead>
<tbody>
<tr>
<td>Abbott et al.</td>
<td>Kevlar 29</td>
<td>0.22-cal notched</td>
<td>556-586</td>
</tr>
<tr>
<td>[55]</td>
<td>Kevlar 49</td>
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<td>480-510</td>
</tr>
<tr>
<td></td>
<td>PBI</td>
<td>0.22-cal notched</td>
<td>426 - 456</td>
</tr>
<tr>
<td>Coskren et al.</td>
<td>Nylon</td>
<td>unknown</td>
<td>615</td>
</tr>
<tr>
<td>[56]</td>
<td>Polyester</td>
<td>unknown</td>
<td>472</td>
</tr>
<tr>
<td></td>
<td>Nomex</td>
<td>unknown</td>
<td>442</td>
</tr>
<tr>
<td></td>
<td>Fiberglass</td>
<td>unknown</td>
<td>274</td>
</tr>
<tr>
<td>Chocron et al.</td>
<td>Kevlar KM2</td>
<td>0.30-cal FSP</td>
<td>628</td>
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<tr>
<td>[53]</td>
<td>Dynema SK65</td>
<td>0.30-cal FSP</td>
<td>550</td>
</tr>
<tr>
<td></td>
<td>PBO</td>
<td>0.30-cal FSP</td>
<td>567</td>
</tr>
<tr>
<td>Carr [50]</td>
<td>Aramid*</td>
<td>0.223-cal sphere</td>
<td>618</td>
</tr>
<tr>
<td></td>
<td>Dynema SK66**</td>
<td>0.223-cal sphere</td>
<td>686</td>
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<tr>
<td>Bazhenov [52]</td>
<td>SVM aramid</td>
<td>30 mm sphere</td>
<td>670</td>
</tr>
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<td>Petterson et al.[146]</td>
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<td>580</td>
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<tr>
<td>Smith et al.</td>
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<td>650</td>
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<tr>
<td>[57]</td>
<td>Polyester</td>
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<td>620</td>
</tr>
<tr>
<td>Wilde et al.</td>
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<td>[58]</td>
<td>Nylon 6/6 - Textile</td>
<td>GPT</td>
<td>345-377</td>
</tr>
<tr>
<td></td>
<td>Nylon 6/6 - Carpet</td>
<td>GPT</td>
<td>271-280</td>
</tr>
<tr>
<td></td>
<td>Nylon 6/6 - Undrawn</td>
<td>GPT</td>
<td>180-200</td>
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<td>Freeston et al.</td>
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<td>[60]</td>
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<td></td>
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<td>442±15</td>
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<td></td>
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<td>335±15</td>
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<td>Chromel wire</td>
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<td>216±15</td>
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<tr>
<td>Claus et al.</td>
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<td>notched dart</td>
<td>610</td>
</tr>
<tr>
<td>[61]</td>
<td>Nylon A07</td>
<td>notched dart</td>
<td>573</td>
</tr>
<tr>
<td></td>
<td>Nylon 728</td>
<td>notched dart</td>
<td>552</td>
</tr>
<tr>
<td></td>
<td>Nylon 300</td>
<td>notched dart</td>
<td>544</td>
</tr>
<tr>
<td></td>
<td>Polypropylene U100</td>
<td>notched dart</td>
<td>527</td>
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<tr>
<td></td>
<td>Polypropylene 301</td>
<td>notched dart</td>
<td>518</td>
</tr>
<tr>
<td></td>
<td>Nomex 430</td>
<td>notched dart</td>
<td>450</td>
</tr>
<tr>
<td></td>
<td>Nylon 330</td>
<td>notched dart</td>
<td>450</td>
</tr>
</tbody>
</table>
The following work shows the results of impacting three different types of yarn (Kevlar KM2, Dyneema SK76, and AuTx®) with three different projectile geometries (round, FSP, and razor blade). The three different material choices have been selected so as to cover the major material types in existence, namely aramid and ultra-high molecular weight polyethylene (UHMWPE). The three projectile types have been chosen so as to cover a wide array of stress conditions that could be induced in the yarn. Finally, rupture surfaces from all three yarn types have been imaged at a range of velocities when impacted by each of the aforementioned projectile nose geometries.

6.3 Experimental

6.3.1 Materials

Three different material types were selected for experiments, namely an aramid homopolymer (Kevlar® KM2), UHMWPE (Dyneema® SK76), and an aramid copolymer (AuTx®). While the former two material types are used heavily in current body armor applications, the lesser known material AuTx® is a relatively new terpolymer-aramid fiber produced by Kamenskvolokno JSC for the Alchemie® group [152]. All materials were in the form of as received-yarn wherein Kevlar KM2 and Dyneema SK76 were provided as an untwisted, or straight tow, while the AuTx® yarn was provided in a pre-twisted configuration. Prior to the onset of the experimental sequence, it was determined that yarns would be tested in the as-received condition, and as such, it must be noted that slight increases in yarn pre-twist have been shown to positively affect longitudinal tensile properties [55,109] due to increased inter-fiber friction. Additionally, Song and Lu tracked transverse wave speeds developed in single Kevlar® KM2 yarns when subjected to slight increases in yarn twist, determining that incremental yarn twist from up to 8/3 twists per inch resulted in an increase transverse wave speed and a reduced transverse wave angle, thereby demonstrating that such increases in yarn twist increase the structural stiffness of the tow [153]. As such, the as-received condition of AuTx® yarn may positively influence yarn longitu-
dinal strength. Due to handling during specimen mounting it is estimated that a max level of user-input twist into the yarn was less than three full rotations per meter. Thus, the user-input pre-twist is considered as negligible, being much less than the optimal strength enhancing condition of 120-240 revolutions per meter as shown by Abbott et al. [55] and Mulkern and Rafternberg [109] for Kevlar yarns.

6.3.2 Quasi-static Longitudinal Tension

Quasi-static tension experiments were performed in a fashion similar to that described in ASTM standards D2256 [111] and D7269 [112], using the slack-start procedure. Single yarns were loaded in a MTS servo-hydraulic universal test system via pneumatic 180° bollards grips and pulled at a strain rate of 0.01/s. Force signals were detected with a 300 lb 1500ASK Interface force transducer and amplified with a Vishay 2310B dual mode amplifier. Displacement was taken directly from the MTS 810 output. Both force signals and displacement signals were simultaneously collected with a Tektronix DPO4032 oscilloscope at a sampling frequency of 2.5 kHz implementing a 1 kHz low-pass filter. As determined from previous study [154], in order to negate yarn slipping, clamping surfaces of the bollard grips were covered with carbon tape (Tad Pella) and yarn ends were coated with athletic rosin powder (Humco). Three sample lengths were tested, having nominal gauge lengths of 444 mm, 666 mm, and 889 mm. Initial elastic modulus, failure strain, and failure stress were all determined from the resulting stress-strain curves.

6.3.3 Transverse Impact Critical Velocity

In efforts to achieve the appropriate velocities needed to cause immediate rupture of single yarns during transverse impact, a 20-mm powder-breech system was employed (New Lenox Ordnance), which can be seen in Figure 6.2. Such a system allows for projectiles of varying diameters via implementation of high-density polyurethane foam sabots (Foam-iT! 26). Sabots were subsequently cast in molds that accept as-
sorted projectile diameter inserts. In the current data set 0.30-cal round, 0.30-cal FSP [131], and razor blade heads were used for impactors. All listed projectiles can be seen in Figure 6.3.

Sabots were placed in the breech of the gun, followed by a small bag of smokeless powder (Hi-Skor 700-x or CFE 223). Ignition of said propellant was promoted by impacting a .30-06 cartridge containing a magnum rifle primer (Remington Large Rifle Magnum Primers 9-1/2M) placed directly behind the powder bag. Primer ignition was induced by impact from a firing pin which itself was impacted by a linear solenoid.
Figure 6.3. Projectiles used for transverse impact experiments. From left to right can be seen the 0.30-cal round, 0.30-cal FSP, and razor blade. All projectiles are housed in polyurethane sabots.
Exit velocity of the projectile was varied via altering the amount of powder placed in the breech or by varying the initial projectile depth into the barrel. In efforts to attenuate muzzle blast created from firing the powder-breech system, a muffler was located immediately downstream of the barrel muzzle. Flight velocity of the incoming projectile was measured after the projectile exited the muffler by a laser-diode system that consisted of three independent laser-diode pairs (shown in Figure 6.2(b)). Velocity was determined via tracking the times at which the projectile interfered with the laser-diode site lines, and the known inter-diode spacing. Incident velocity was also verified with high-speed imaging and timing of sample impact. An example of projectile exit velocity as a function of smokeless powder mass can be seen in Figure 6.4.

Yarns were hung between two load cells (Kistler 9712) and were attached via aluminum clamps. In efforts to ensure perpendicularity of the yarn to the bore shot line, a plumb-bob was used to in efforts to achieve a vertically positioned yarn and a bubble level was used to achieve a horizontal barrel position. With respect to the projectile flight path, the yarn position was centrally located within the middle of the barrel shot line by alignment with a muzzle loaded laser bore-site. Yarns were strung with minimal tautness ensuring negligible preload; pre-loads were kept below
1 N. The yarn length was 0.85 m, although there was an offset made for the shot location so as to measure longitudinal wave speeds, which is subject of future report. Meaningful yarn deformation was tracked only for times before the interaction of the the reflected longitudinal stress wave and the primary transverse wave. Similar to the quasi-static experiments, user imposed yarn pre-twist was kept below three revolutions per meter.

Initially, a prediction of velocity was used to pre-determine an estimated time of projectile-yarn impact delayed from the passage of the projectile in front of the third laser-diode. Although functional, such a method required low framing rates (\(\sim 40K\) fps) in order to ensure reasonable success of capturing the impact event, especially when firing at the lower velocities required for the razor blade experiments. While feasible, such low recording rates left image quality quite poor due to the inherently long exposure times. In efforts to increase framing rate while ensuring capture of the time window of interest, post impact triggering was utilized via the rolling onboard image storage capability of the high-speed camera system. Due to the consistent velocity of the longitudinal stress wave (\(c\) in Equation 6.1 and Figure 6.1), a fast response quartz transducer (Kistler 9712) was used to detect the stress wave arrival to a yarn-end gripping clamp. The subsequent transmitted load generated in the wake of the longitudinal wave front was thus captured. Such a trigger method was proven to be of extreme use in the current data set, allowing for a high success rate of capturing the impact event.

From each transverse impact experiment, determination of the ensuing yarn behavior was tracked via the resulting high-speed image sequence, with examples of such impact events shown in Figure 6.5. Upon visual inspection, determination of the aforementioned critical velocity (wherein the yarn behavior transitions from transverse wave front development to no significant wave propagation due to immediate failure) can be separated into three velocity regions of interest, specifically, zero yarn failure, partial yarn failure, and full yarn failure. An example of a velocity nearly at the transition zone between no yarn failure and partial yarn failure can be seen
in Figure 6.5(a), being defined by development of a transverse wave front and minimal filament failure. Similarly, the transition velocity between partial yarn failure and full yarn failure is shown in Figure 6.5(c), wherein nearly the entire yarn fails, thereby yielding transverse wave propagation development for only a few filaments. Within the transition zone, a fraction of the filaments fail upon impact, while the remainder exhibit transverse wave front development. An example of such a partial failure can be seen in Figure 6.5(b). Although visually differentiable, it must be noted that determination of the transition regions is decidedly qualitative in nature. Post analysis of wave front development and filament failure for a specific shot, the result was labelled to be within one of the three aforementioned regimes, depending on the presence of filament failure and transverse wave development. Once a comfortable number of data points encircled both boundaries dividing the three regimes, specific velocities were selected to demarcate the transition between: (1) no filament failure to slight filament failure and (2) nearly all filament failure and complete filament failure. Example of this procedure can be seen in Figures 6.6-6.8. Said transition demarcations are averaged between the velocities of two adjacent shots, slightly below and slightly above the appropriate transition characteristic, having not more than a velocity difference of \(\sim 30\) m/s (although most regions differ by less than 20 m/s).

### 6.3.4 Transverse Impact Fracture Surfaces

Failed yarn ends were collected after each experiment so as to provide fracture surfaces for subsequent imaging and analysis. Samples were selected from four relative impact velocities throughout the critical velocity transition region (slightly below the transition region, near the lower end of the transition region, near the upper end of the transition region, and above the transition region) for all three yarn types impacted with each of the three projectile heads. Effort was taken to select yarns that displayed little to no interaction with the following projectile containment assembly. Fractured yarn ends were attached to an aluminum substrate with carbon tape (Ted Pella) and
Figure 6.5. Image sequence of a razor blade impact into Kevlar® KM2 near the (a) lower limit (b) middle, and (c) upper limit of the transverse impact critical velocity. Razor blade projectile velocity is measured at (a) 141 m/s, (b) 196 m/s, and (c) 318 m/s. Inter-frame spacing is (a) 3.97 μs, (b) 3.97 μs, and (c) 3.91 μs.
then coated with AuPd using an SPI-Module sputter coater. Coated samples were then imaged via scanning electron microscopy (SEM) with either an XL40 or Hitachi S-4800 using accelerating voltages of 5kV or 30kV and beam currents of 310-340 μA, respectively.

6.4 Results/Discussion

6.4.1 Quasi-static Longitudinal Tension

Yarns were pulled in quasi-static tension in a fashion similar to ASTM standards D2256 [111] and D7269 [112] so as to ascertain a representative failure strain needed for input into the analytical solution described in Equations 6.1-6.5. All three material types were pulled at a strain rate of 0.01 s⁻¹ using gauge lengths of 444, 666, and 889 mm. Such long gauge lengths were chosen so as to minimize gripping end effects in the bollards, while still examining gauge length effects. Failure strain, failure stress, and initial elastic modulus were determined from each testing condition with resulting properties listed in Table 6.2. Elastic modulus and failure strain input values for Equations 6.1-6.5 were determined via averaging results from all three gauge lengths, yielding values of 127.48 GPa and 2.68%, 99.49 GPa and 2.57%, and 131.04 GPa and 2.91% for Dyneema® SK76, Kevlar® KM2, and AuTx®, respectively. No demonstrative effect of gauge length was detected for the range of lengths within current interest. As previously described, it has been shown that frictional effects from a slight twisting of yarn can yield an increase in yarn tensile strength [38,55,92,109,155], so effort was made to ensure minimal yarn twist (less than three revolutions per meter) for the experiments yielding the response described in Table 6.2. It is important to note that such a nominal zero twist condition displays a noticeable reduction in tensile strength and failure strain values found in literature as compared to that shown from manufacturers. It is assumed that such a difference is due to the lack of twisting in the reported literature results. For example, untwisted Kevlar® KM2 yarn has been reported in literature to exhibit a quasi-static failure stress of
Table 6.2. Yarn quasi-static longitudinal properties. Note: Linear density of Dyneema SK76 and Kevlar KM2 was taken directly from the yarn bobbin. *Linear density of the AuTx® yarn was not provided by the supplier; it was measured in house and further information can be found in [154].

<table>
<thead>
<tr>
<th>Fiber Type</th>
<th>Linear Density (denier)</th>
<th>Gauge Length (mm)</th>
<th>Failure Strain (%)</th>
<th>Failure Stress (GPa)</th>
<th>Elastic Modulus (GPa)</th>
<th>Number of Samples</th>
</tr>
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<tbody>
<tr>
<td>Dyneema® SK76</td>
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<td>444</td>
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<td>2.44±0.08</td>
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<td></td>
<td>889</td>
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<td>2.46±0.09</td>
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<td>AuTx®</td>
<td>275*</td>
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<td>2.86±0.12</td>
<td>3.09±0.14</td>
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<td>129.44±1.07</td>
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</table>

2.87 GPa (GL = 24 mm, SR = 10⁻⁵ s⁻¹ - [101]), ~2.7 GPa (GL = 381 mm, SR = 8.34x10⁻³ s⁻¹ [102]), and 2.66 GPa (GL = 500 mm, SR = 4x10⁻³ s⁻¹ - [109]), while the latter study reports a strength of 3.36 GPa when KM2 is twisted to 3.4 revolutions per inch (Note: gauge length and strain rate are abbreviated as GL and SR, respectively). Similarly, literature has reported quasi-static tensile strength and failure strain values for Dyneema® SK76 yarn to be 2.33 GPa and 2.28% (GL = 5-20 mm, SL = 10⁻⁴ s⁻¹ - [156]), while the failure stress and rupture strain for Dyneema SK75 and SK78 are by the manufacturer to be 3.3-3.9 GPa and 3-4%, respectively [140].

6.4.2 Transverse Impact Critical Velocity

Determination of no/partial/full yarn failure is depicted in Figures 6.6, 6.7, and 6.8 for all transverse impact experiments into Dyneema® SK76, Kevlar® KM2, and AuTx®, respectively. Additionally, the upper and lower bounds of the transition regions for all three fiber types can also be found in Table 6.3. For Figures 6.6-6.8, razor blade, FSP, and round data are demarcated by light grey triangles, black squares, and dark grey circles, respectively. Vertical solid lines have been drawn on each plot signifying both the beginning and end of the partial yarn failure transition.
Figure 6.6. Resulting data set for transverse impact experiments into Dyneema® SK76 yarn using round, FSP, and razor blade projectiles. Solid vertical lines represent estimated transition velocities from the current data set while the dotted vertical lines represent transition velocities from Chocron et al. [53], who tested Dyneema® SK65. Dash dot lines represent predicted critical velocities for each projectile head [130], and the dash double dot line represents the critical velocity predicted from quasi-static longitudinal yarn experiments shown in Table 6.2.

zones. Although not shown in Figures 6.6-6.8, within this transition zone it is possible to see an evolution in the ratio of failed and intact filaments, but in the current work, no effort has been made to signify such an evolution as it was determined to be much too subjective. Critical velocities as predicted from quasi-static experimental results shown in Table 6.2 have been overlaid in Figures 6.6-6.8 for each type of fiber. Additionally, critical velocities are predicted for Dyneema® SK76 and Kevlar® KM2 via quasi-static transverse loading experiments from Hudspeth et al. [130] using all three projectile geometries, and are overlaid in Figures 6.6 and 6.7, respectively.

Lower to upper velocities defining the critical velocity transition regime exhibited by Dyneema® SK76 due to impact via razor blade, FSP, and round impactors were found to be 190-315 m/s, 450-690 m/s, and 505-750 m/s, respectively. While the razor blade transition velocity was clearly below the transition region of both the
FSP and round projectiles, the latter two impactors showed only a slight difference. Such a small variation, although initially unexpected, may be due to the radius of curvature of the FSP, which is discussed below. Similar to the geometry of the current experiments, Chocron et al. [53] impacted Dyneema® SK65 yarns with a 0.30-cal FSP, resulting in a critical velocity of 517-583 m/s, which lies within the FSP transition region for the current SK76 experiments. It is also important to note that the presumed critical velocity determined for spherical impact of Dyneema® SK66 was found by Carr to be 686 m/s (Carr, 1999), which lies within the round projectile experimental transition regime of SK76 found in the current study.

In addition to the experimental transverse impact critical velocity transition regime, the predicted critical velocity from single filament quasi-static transverse loading data for Dyneema® SK76 (Hudspeth et al., 2015b) has also been overlaid on Figure 6.6, being 612 m/s, 1022 m/s, and 1235 m/s for the razor blade, FSP, and round, indenters, respectively. For all three projectile heads, the single-fiber quasi-static estimation of the transverse critical velocity over predicts the actual velocities found from yarn experiment, which is not completely unexpected, as single filaments generally exhibit a larger tensile failure strain and rupture strength as compared to yarn [55,154]. Indeed, such a difference in tensile failure strain is seen for Dyneema SK76, being ~3.5% [130,157] and ~2.7% (Table 6.2) for single fiber and yarn, respectively. As such, if one were to normalize all predicted single fiber failure strains from the transverse quasi-static experiments using a similar ratio to that found from longitudinal tension experiments (i.e. 2.7/3.5), estimated critical velocities would be on the order of 510 m/s, 840 m/s, and 1020 m/s, for the razor blade, FSP, and round indenters, respectively. Purely a hypothetical reduction, such velocities do fall closer to the experimental transition regimes, although still greater than that which is seen in experiment. Regardless, the experimental transverse critical velocity regimes for Dyneema® SK76 are clearly a function of the projectile nose geometry and all experimental results are vividly below the typically assumed pure tension longitudinal failure criterion.
Figure 6.7. Resulting data set for transverse impact experiments into Kevlar® KM2 yarn using round, FSP, and razor blade projectiles. Solid vertical lines represent estimated transition velocities from the current data set while the dotted vertical lines represent transition velocities from Chocron et al. [53]. Dash dot lines represent predicted critical velocities for each projectile head (Hudspeth et al. 2015b) and the dash double dot line represents the critical velocity predicted from quasi-static longitudinal yarn experiments shown in Table 6.2.
The resulting critical velocity transition regions for Kevlar® KM2 are plotted in Figure 6.7 for each of the projectile heads of interest, namely razor blade, FSP, and round, yielding lower to upper velocities of 145-310 m/s, 480-645 m/s, and 540-700 m/s. Again, as found with the SK76 results, the razor blade projectile demonstrated a drastic reduction in the critical velocity regime as compared to the FSP and round projectile heads. Interestingly, the FSP and round projectiles demonstrated reasonably similar critical transition velocities, most likely due to the radius of curvature of the FSP geometry, which is discussed below.

Estimated critical velocities from quasi-static transverse loading experiments performed by Hudspeth et al. [130] are overlaid on Figure 6.7 for Kevlar® KM2, being 388 m/s, 661 m/s, and 782 m/s for razor blade, FSP, and round projectiles, respectively. As compared to the results from Dyneema® SK76, Kevlar® KM2 predicted critical velocities for each projectile head are much closer to that demonstrated in actual transverse impact, being only slightly greater than the upper limit of each respective transition regime. Furthermore, as with the Dyneema® SK76 data, if the Kevlar® KM2 results are normalized by the longitudinal fiber [130,157] to yarn (Table 6.2) failure strain ratio (i.e. 2.6/3), resulting predicted critical velocities are reduced to 343 m/s, 584 m/s, and 693 m/s, for the razor blade, FSP, and round projectiles, respectively. Again, although purely hypothetical, such a fiber to yarn normalization does render critical velocity values within the experimental transition regime for both FSP and round projectiles and near to the upper boundary of the results displayed by impact with the razor blade.

The experimental critical velocity of Kevlar® KM2 determined by Chocron et al. [53] can be seen overlaid in Figure 6.7, sitting at the top, yet inside of the range determined from the current experiments. Reasoning for such positioning of the Chocron et al. [53] critical velocity is due to their definition of yarn failure; a lack of any transverse wave development due to rarefaction from filament rupture is defined as their yarn critical velocity [158]. Such a delineation is identical to the upper transition definition of the current study. That said, the results from Dyneema® in
Figure 6.6 show the velocity range from Chocron et al. to be closer to the bottom of the data set from the current experiments, but it must be noted that they tested SK65, which presumably has a reduced strength and strain as compared to SK76, as described by trends in the manufacturer data sheet [140]. Thus the upper end of the transition region for SK65 very well may lie somewhat below that of the SK76 material. Further verifying the transition regions in the current data set, it is also useful to mention the critical velocities described by both [50,55]. The former work impacted both Kevlar® 29 and Kevlar® 49 with a notched saddle projectile, yielding critical velocities of 556-586 m/s and 480-510 m/s, respectively, while the latter shot Kevlar® 129, Kevlar® KM2, and Twaron® CT via sphere, finding a presumed averaged critical velocity of 618 m/s. While both studies are using a myriad of different aramid materials, when compared to the results from KM2 listed in Figure 6.7, both the Abbott et al. and Carr studies exhibit aramid critical velocities lying within the current experimental transition region found for Kevlar® KM2 when impacted with the round head projectile. It is also interesting to note the increase in critical velocity exhibited by newer aramid materials, as evidenced in the improvement in critical velocities from older Kevlar® 29 and 49, to the newer KM2, is presumably due to manufacturing advancement.

For Kevlar® KM2, although the critical velocity predicted from each quasi-static loading condition is above that which is seen from actual impact, the prediction is much closer to the true experimental critical velocity for both the razor blade and FSP than the prediction from single yarn longitudinal quasi-static experiments, which is estimated at 727 m/s from Table 6.1. As with the results from Dyneema® SK76, the quasi-static yarn tension experiment is found to predict a critical velocity below that which is seen from the round projectile head, suggesting that single fiber experiments result in a greater tensile strain than full yarn experiments. Again, such a finding is not uncommon [55,154]. Furthermore, while the transverse impact of yarn with a round indenter does indeed show a critical velocity regime lower than both the predicted critical velocities from yarn longitudinal tension experiments and single
Figure 6.8. Resulting data set for transverse impact experiments into AuTx® yarn using round, FSP, and razor blade projectiles. Solid vertical lines represent estimated transition velocities from the current data set. The dash double dot line represents the critical velocity predicted from quasi-static longitudinal yarn experiments shown in Table 6.2.

fiber transverse loading experiments, the upper end of the transition region is not drastically lower than both predictions. Finally, it is again important to note that for both the Dyneema® SK76 and Kevlar® KM2 yarns, there is clearly an effect of projectile nose geometry on the critical velocity transition regime and all critical velocity regimes are most definitely below the predicted critical velocity from yarn longitudinal tension experiments.

The lesser known material AuTx® was also subjected to a range of transverse impact experiments, although yielding trends slightly different from both Dyneema® SK76 and Kevlar® KM2. While transverse impact razor blade experiments showed by far the lowest critical velocity regime, being 180-340 m/s, the transition regime for the round indenter brackets the FSP data, rather than being shifted upwards to a higher velocity. The round indenter has a critical velocity regime of 475-805 m/s while the FSP sits inside of this regime, having a span of 540-720 m/s. Although reasoning for this odd shift in the critical velocity transition regions for the FSP and
Table 6.3. Transition velocities found during transverse impact of Dyneema® SK76, Kevlar® KM2, and AuTx®. Also listed are quasi-static predictions from Hudspeth et al. [130], and classical solution critical velocities generated from Table 6.2 and Equations 6.1-6.5. Note: *Quasi-static prediction velocities have been determined from experimental failure strains found in Hudspeth et al. [130]. **Classical velocity predictions have been derived from failure strain and elastic modulus averages found in Table 6.2.

<table>
<thead>
<tr>
<th>Fiber Type</th>
<th>Projectile Type</th>
<th>Lower Critical Velocity (m/s)</th>
<th>Upper Critical Velocity (m/s)</th>
<th>Lower Critical Strain (%)</th>
<th>Upper Critical Strain (%)</th>
<th>Quasi-static Prediction (m/s)*</th>
<th>Classical Velocity Prediction (m/s)**</th>
</tr>
</thead>
<tbody>
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<td>Dyneema® SK76</td>
<td>Razor</td>
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<td>315</td>
<td>0.27</td>
<td>0.54</td>
<td>612</td>
<td>1030</td>
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<tr>
<td></td>
<td>FSP</td>
<td>450</td>
<td>650</td>
<td>0.87</td>
<td>1.54</td>
<td>1022</td>
<td></td>
</tr>
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<td>Round</td>
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<td>750</td>
<td>1.01</td>
<td>1.73</td>
<td>1235</td>
<td></td>
</tr>
<tr>
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<td>310</td>
<td>0.29</td>
<td>0.81</td>
<td>388</td>
<td>727</td>
</tr>
<tr>
<td></td>
<td>FSP</td>
<td>480</td>
<td>645</td>
<td>1.46</td>
<td>2.18</td>
<td>661</td>
<td></td>
</tr>
<tr>
<td></td>
<td>Round</td>
<td>540</td>
<td>700</td>
<td>1.71</td>
<td>2.43</td>
<td>782</td>
<td></td>
</tr>
<tr>
<td>AuTx®</td>
<td>Razor</td>
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<td>340</td>
<td>0.3</td>
<td>0.7</td>
<td>N/A</td>
<td>975</td>
</tr>
<tr>
<td></td>
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<td>1.93</td>
<td>2.25</td>
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</tr>
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<td></td>
<td>Round</td>
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<td>805</td>
<td>1.1</td>
<td>2.25</td>
<td></td>
<td></td>
</tr>
</tbody>
</table>

Round projectile heads is most likely due to the FSP projectile radius of curvature (discussed below), it is reasonable to point out that Walker and Chocron [54] do state that the explained bounce phenomenon does apply to rounded projectile heads as well as flat-faced projectiles. Even though such a failure process is not agreed upon by the current authors, it is possible that the AuTx® fiber is more adept to housing shear loads [157], thereby allowing for alternative modes of failure to occur, not being as greatly affected by local multi-axial stress states. It is also important to note that as with the previously mentioned yarns, prediction of critical velocity of AuTx® yarn from quasi-static tensile properties, which is 975 m/s, is much greater than the aforementioned critical velocity transition regimes displayed from impacts of all three projectile heads.

It is now of importance to discuss the minimal difference in transition velocities between the FSP and round projectiles, which was demonstrated for all three fiber types. This behavior, which became of interest after completion of the impact experiments, is believed to have occurred due to a blunting of the FSP radius of curvature after heat treatment. As compared to previously performed quasi-static experiments [130],
projectiles used in this experimental set were sandblasted post heat treatment, as is common in machining practice. Such a blunting chamfered the FSP corner geometry. Mounting, cutting, and polishing the FSP cross section yielded a radius of curvature of 87 m, which is shown in Figure 6.9(a). In order to assess this effect of curvature radius on transition velocity, five projectiles were machined excluding sandblasting, which were measured to have a curvature radius of 6.5 m, as shown in Figure 6.9(b). Sharpened FSPs were then shot into Kevlar® KM2, with a focus on determining the upper bound of the critical transition velocity. As seen from the resulting shot data shown in Figure 6.10, the sharpened FSP yielded a transition velocity of 600 m/s as compared to the 645 m/s for the abraded FSP. This reduction in transition velocity is to be expected, as it is believed the smaller radius of curvature of the non-sandblasted projectile can create a greater stress concentration around the corner of the projectile edge, further demonstrating the notion that projectile geometry highly governs failure in yarn during transverse impact. It is thus postulated that additional experiments using sharp FSP corner geometries would reduce the entire FSP transition regime for all fiber types tested.
Figure 6.10. The upper velocity of the critical velocity transition region was determined for Kevlar® KM2 using the sharpened FSP, showing a reduction in value when compared to the non-sharpened FSP projectile head.
Although the subject of work in the immediate future, it is of importance to briefly frame the consequence of the described transverse critical velocity, and ultimately how it may affect full fabric performance. As previously stated, impact of full fabric systems will result in local ‘inelastic’ failure at sufficiently high impact velocities [7,9,11,49]. Akin to transverse impact of single yarns, such an ‘inelastic’ failure is promoted when yarn rupture occurs almost immediately upon impact, removing any substantial waveform development and ultimately, any legitimate energy dissipation (e.g. longitudinal elastic strain energy, longitudinal kinetic energy, and transverse kinetic energy) from the layer(s) of fabric demonstrating immediate local failure. As such, the linear properties displayed by yarns within high-performance fabric undergoing local rupture are rendered nearly useless in halting the incoming projectile and can even be replaced by less expensive surrogate material [49]. This behavior is not limited solely to extremely overmatched impact conditions, as this ‘inelastic’ failure mode is promoted below the ballistic limit of a full system for plies located in the frontward region of the panel initiating projectile contact [159]. Although below the ballistic limit of the entire pack, these frontward panels see an impact velocity much above their own piecewise ballistic limit. As such, it is of extreme interest to understand the immediate yarn rupture phenomena, as it inherently governs the initial failure demonstrated in the front panels within a fabric system impacted near the system ballistic limit, or failure of the entire pack when shot in an overmatched condition.

At this point it is now of interest to discuss the oft-quoted Cunniff equation, which does a sufficient job at predicting the impact behavior of aramid fabric systems [10]. The beauty and power of this non-dimensional equation, defined as \( U^{1/3} = \frac{\sigma \varepsilon}{2 \rho \sqrt{E / \rho}} \), lies in its capability to predict possible ballistic limit velocities of a candidate fabric material if one has performed a longitudinal tension experiment and possesses previously performed ballistic limit data of a different fabric system. \( E, \rho, \sigma, \varepsilon \) are defined as the elastic modulus, fiber density, longitudinal rupture force, and longitudinal failure strain. One can think of this equation as the product of two parts, namely, the specific
energy to yarn rupture and the longitudinal wave speed. Ideally, one would want a fiber that absorbs as much energy as possible before failure, while being able to move that energy away from the impact sight as quickly as possible [60]. Although this parameter is powerful and works quite well for aramids, predictions for other material types do show some variation. It is suggested that such variation may inherently derive from the nature of failure from one material type to another. For example, Nylon 6/6 has been shown to have a ballistic critical velocity of 650 m/s [119], being quite similar to aramid and UHMWPE material, yet Nylon specific toughness and longitudinal wave speed are much lower than that of modern high-performance fibers, which as suggested by the Cunniff parameter, reduces the ballistic limit [10,150]. In contrast, glass fiber has been shown to possess a Cunniff parameter similar to modern polymeric high-performance fiber when using dynamic tensile properties, yet the ballistic limit of glass systems does not fall along the failure envelope described by the Cunniff equation [10], likely due to its brittleness in transverse loading. It is thus proposed that accurate prediction of a candidate fiber material must possess an additional consideration, being governed by transverse properties that are teased out from the current analysis of the critical velocity. Ideally, a fiber material should not only possess superb longitudinal mechanical properties described by the Cunniff equation, but should also contain a high degree of transverse/shear strength, possibly resulting in an increased critical velocity. As an example, it is suggested that the superior ballistic limit of fabric panels composed of M5 fiber as described by Cunniff et al. [122] was due to the presence of intermolecular transverse hydrogen bonding. As the ballistic limit data showed a 25-40% performance increase over that expected from the Cunniff equation, it is believed that such a rise could be captured from a preliminary determination of critical velocity transverse impact experiments into single yarns, or even single fiber quasi-static transverse loading experiments described by Hudspeth et al. [139], which are both much more cost effective experiments than shooting entire ballistic panels, as much less material is required. Of course impact of full fabric systems is ultimately obligatory for full body armor panels before they
are approved for use in armor applications, but such a tool could be used to qualify candidate materials early in the research phase that are prohibitively expensive to produce.

6.4.3 Transverse Impact Fracture Surfaces

As described by Hearle [38] and [37], an understanding of fiber failure can be elucidated by analyzing the resulting single filament fracture surfaces. Thus, in efforts to gain additional insight into the failure modes exhibited during yarn transverse impact, fracture surfaces have been recorded from yarns impacted at a variety of velocities using all three aforementioned projectile nose geometries, namely razor blade, FSP and round. All three yarn types (Kevlar® KM2, Dyneema® SK76, and AuTx®) have been investigated at specific velocity regions bracketing the respective transition regimes for each projectile nose geometry. Impact velocities are grouped into four categories for each projectile nose geometry: below the lower boundary of the transition region, near the bottom of the transition region, near the top of the transition region, and above the transition region. Representative fracture surfaces using all three projectile nose geometries at each of the four described velocities from Dyneema® SK76, Kevlar® KM2, and AuTx® can be seen in Figures 6.11, 6.12, and 6.13, respectively. It should be noted that fracture surfaces gathered from impact conditions wherein the yarn yarn was shot at a velocity above the respective critical velocity originate from the projectile-yarn contact site wherein failure occurs before any boundary wave reflection. Micrographs taken from impact conditions wherein partial failure occurs have been centered about the most common failure surface; thus higher impact velocities result in more filaments exhibiting immediate rupture and lower impact velocities inevitably fail at later times which may be subjected to multiple wave interactions. Finally, fracture morphologies taken from impact conditions wherein the striking velocity is below the critical velocity transition region most definitely fail due to multiple wave interaction, and thus fracture surfaces have been
Figure 6.11. Representative fracture surfaces exhibited by Dyneema SK76 fibers when impacted below, within, and above the critical transition velocities. Note the local failure seen from razor blade impact. Surfaces from both round and FSP impact show melting, either caused by the failure process or filament snap-back.

taken from the point of failure, which is either at a clamp (due to stress increases from rigid boundary conditions) or from the projectile-yarn contact site (due to wave reverberations and stress concentration development from the contact geometry).
According to the morphologies presented in Figure 6.11, Dyneema® SK76 is dominated by cutting or short range fibrillation, with a trend towards local cutting as the projectile nose geometry becomes sharper. Additionally, all fracture surfaces appear to exhibit a good deal of melting similar to that found from Iremonger [13] and Carr [50], being most definite with the round projectile head. Additionally, this melting observation increases with increasing impact velocity. Due to the low melting temperature of UHMWPE, such a fracture surface may be caused by the snap-back of the filaments that undergo a heavily tensile dominated failure mode. The snap-back phenomenon is generally described by the release of the elastic energy stored in individual fibers immediately post rupture [38]. Such a failure process has been described by Hearle who notes that rapid plastic flow at high velocity may produce heat that causes the fiber to melt [38]. It is also possible that the perceived melting process is solely promoted by contact of filaments with the projectile. Smearing/flattening is also present for some of the fractured surfaces impacted within the transition regions by FSP or round indenters which has been previously seen during both low [128,129] and high-rate [13,15] transverse loading. With regards to the fracture surfaces produced by razor blade impact, it can be seen that localized shear cuts have formed being flat in nature and containing some degree of smearing and melting.

As compared with Dyneema® SK76, the rupture morphologies presented by Kevlar® KM2 demonstrate a greater dependence on the projectile nose geometry and impact velocity. As can be seen in Figure 6.12, when impacted via razor blade, filaments demonstrate localized cutting with a slight degree of fibrillation for all examined impact velocities, quite similar to the fracture surfaces seen from Shin et al. [128], Mayo and Wetzel [129] and Hudspeth et al. [130]. At low impact velocities (below the transition region and at the bottom of the transition region), filaments also exhibit a large degree of smearing. Above the transition region, impact with razor blades still promotes localized failure but axial splitting also becomes evident. When impacted with the rounded indenter, fibrillation and axial splitting failure dominates for all analyzed fracture surfaces throughout the entire velocity range of interest, although
Figure 6.12. Representative fracture surfaces exhibited by Kevlar KM2 fibers when impacted below, within, and above the critical transition velocities. Note the local failure seen from razor blade impact and the high degree of fibrillation seen from round impact. Filaments recovered from FSP impact suggest a more localized failure when impacting above the critical velocity transition zone.
there does seem to be a slight trend towards shorter fibril lengths when increasing from lower to higher impact velocities. Interestingly, filaments impacted by the FSP show long range fibrillation failure at low impact velocities, while higher impact velocities promote a more localized failure, although fibrillation is still present. Such a trend in fracture surface morphology is similar to that demonstrated in single fiber quasi-static experiments presented in earlier work [130] wherein failure localization was seen to occur at increased loading angles.

AuTx®, similar to Kevlar® KM2, reveals a localized shearing/cutting failure when impacted by razor blades at all velocities of interest, which is demonstrated in Figure 6.13. Additionally, fracture surfaces developed from razor blade projectiles impacting above the transition velocity demonstrate axial splitting, again being quite similar to the response exhibited from Kevlar® KM2. Filaments impacted by the FSP and round indenters both show a high level of fibrillation, but in contrast with the KM2 fiber, a variation from long-range to short-range failure is not seen, as fracture surfaces look nearly identical for all filaments for the entire range in velocity, corroborating the similarity in transition velocities exhibited by AuTx® when impacted by FSP and round projectiles.

In general, more than one fracture mode is observed from each fiber type, such as the slight degree of clean localized cutting or long range fibrillation exhibited by Kevlar® KM2 when impacted with razor blade or round projectiles, respectively. It is suggested that this variation in failure mode is caused by the actual contact condition between the yarn and the indenters during the impact event. Of course these experiments were performed on full yarns, which are inherently composed of numerous single filaments, however, when the yarns are subjected to transverse impact loading with various indenter shapes, some fibers break prior to others depending on which part of the yarn is impacted first or is at the sharpest point of projectile contact [91]. For example, the presence of both localized cutting/shearing and fibrillation occurring in Kevlar® KM2 yarn from one specific shot may be a consequence purely of position within the tow. Cut fibers may be positioned directly beneath the footprint
Figure 6.13. Representative fracture surfaces exhibited by AuTx® fibers when impacted below, within, and above the critical transition velocities. Note the local failure seen from razor blade impact and the high degree of fibrillation seen from FSP and round impact.
of the projectile while subsequent filaments that express a degree of fibrillation may be impacted with a dulled projectile, either from projectile plastic deformation or via ‘front-line’ filament shielding. Regardless, analysis of the current fracture morphologies among the various yarn types does provide a more in-depth recognition of failure behavior due to the assorted impact events.

### 6.5 Dynamic Finite Elements Analysis

In order to determine the stress state developed in a single filament when subjected to transverse impact loading, an explicit FE model has been generated, containing a similar half-symmetry geometry to that used in the quasi-static analysis shown in Section 5.5, but with a much shorter fiber length as the initial transient behavior is of immediate interest. The same transversely isotropic material model used for the quasi-static analysis from Section 5.5 was implemented for the dynamic model and contact conditions were again set to a global tangential friction coefficient of 0.2. Simulations were performed in Abaqus Explicit employing roughly 500,000 trigonal elements on an 8 core CPU with 32 Gb or memory. Typical wall clock run times were 10-60 mins, depending on the simulation duration and mesh coarseness.

Color intensity distribution plots of a FSP projectile impacting a single filament is shown in Figure 6.14. Specifically, 6.14(a), 6.14(b), and 6.14(c) depict S11, S22, and S12 stress states developed near the projectile contact site at a time $\sim 25$ ns post initial contact. In each of the color intensity plots, numbered markers have been used to describe regions of max stress concentrations, being very similar to the locations used for quasi-static analysis as described Figure 5.12 of Section 5.5. Zone 1, which is oriented on the upper region of the filament located above the projectile, is described by a large longitudinal tensile stress due to bending of the filament. Although this bending stress would be reduced by transverse plastic deformation, high local tensile stresses may still be of great enough magnitude to initiate failure. Zone 2, which contains the filament region close to the immediate contact site just
Figure 6.14. Color intensity plots of (a) S11, (b) S22, and (c) S12 stress profiles around the edge of an FSP projectile contact site loaded due to a 500 m/s strike velocity. 1-, 2- directions are locally oriented along the fiber longitudinal and transverse directions, respectively. Legend units are in N/μm².
above the projectile head, experiences a coupled compressive longitudinal stress due to bending and a compressive transverse stress due to transverse compression of the fiber from the projectile contact. Although such a stress state is quite large in magnitude, which is shown in the stress history plot in Figure 6.16, it is important to remember that these stresses are compressive in nature, and as such, most likely cause very little damage to the filament material and likely is not the region of failure initiation.

Finally, zone 3, which is located in the vicinity of the initiated transverse stress wave, shows a highly localized shear stress and transverse tensile stress. Although such a local deformation geometry may seem inappropriate upon first glance, it is important to point the reader to a post-mortem fracture surface found for a Dyneema SK62 filament shot with a razor blade at 636 m/s (Figure 6.15); local stresses during the impact event appear to have caused a local shearing that resulted in the formation of massive kink bands. It is thus postulated that these kink bands may have formed from local shear in the filament, as demonstrated in Figure 6.14(c).

No effort has been made to decidedly determine which zone does indeed initiate failure, but as previously discussed, it is again suggested that failure most likely occurs in zone 1 or zone 3, but this is purely conjecture, as current imagine methods to not allow for fine enough resolution to make out the actual failure zone. That said, as will be discussed in Chapter 10, emerging techniques may indeed provide the necessary temporal and spatial resolutions to probe the local failure process via experiment. It is again important to note that previous work on single fibers have suggested that filaments always fail via shear mechanisms, due to the extremely high stress required to scission the covalently bonded chains oriented along the fiber longitudinal direction [31, 71, 76, 81, 83]. As such, the local shear stresses induced in zone 3 could indeed promote material failure. It is also important to reiterate that this model does not contain plasticity, which would seemingly decrease the bending stresses felt by the filament in zones 1 and 2.

An example of the maximum magnitude stress profile histories developed from impact of a projectile into the single filament can be seen in Figure 6.16. Specifically,
Figure 6.15. SEM of single Dyneema\textsuperscript{®} SK62 shot with a razor blade traveling at a velocity of 636 m/s. Note the cutting failure incurred from the local razor blade contact conditions. Additionally, note the red markers pointing to a heavy degree of kinking, which looks to run through the entire filament thickness, and bears a striking resemblance to the local shearing/kinking depicted in the FEA simulation shown in Figure 6.14.
Figure 6.16. Time history of stresses developed within a single filament subjected to transverse impact loading. Filament properties are identical to those shown in Table 1 and the impact velocity used is 500 m/s.

6.16 demonstrates the impact condition of a FSP impacting a single filament at 500 m/s, with the plotted stress profiles spanning times from initial contact to the point at which stresses sufficiently equilibrate from the initial transient variations.

Similar stress profiles for the dynamic response are developed as compared to the $\sim 30^\circ$ quasi-static analysis, as an impact of 500 m/s into the fiber generates an impact angle of 28°. It is important to note that long range validation of the FE model can be made via comparison with the analytical solution shown in Equations 6.1-6.4 and the resulting comparison can be seen in Table 6.4.
Table 6.4. Long-range comparison between analytical model and FEA solution for single filaments subjected to transverse impact.

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<tr>
<th>Model Type</th>
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<th>FEA</th>
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<tr>
<td>Transverse Wave Speed (m/s)</td>
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</table>
6.6 Conclusions

Ultimately, it is the goal of this work to not only understand at what impact condition a yarn will fail, but more importantly, to stem the development of a relevant failure criterion that can be utilized in full armor system modeling. As such, single yarns have been impacted at a variety of velocities in efforts to determine the transverse critical velocity wherein no substantial transverse waveform development incurs due to the premature failure of the yarn at the impact site. Kevlar KM2, Dyneema SK76, and AuTx® yarns have been impacted with three different projectile geometries, namely razor blade, FSP, and round. All three yarn types were clearly affected by the projectile head geometry, with the razor blade impact results showing the lowest critical velocity for all yarns. For Kevlar KM2 and Dyneema SK76, the FSP geometry showed a slight decrease in critical velocity from the round head, while no discernable delineation of critical velocity could be made for the AuTx® yarn when impacted by the FSP and round projectiles. Post transverse impact experiments, ruptured yarn ends were imaged in order to determine the mode of local filament failure. AuTx® and Kevlar KM2 yarns impacted with the razor blade geometry were shown to exhibit a cutting fracture surface with a slight degree of axial splitting for fibers impacted above the transition regime. Additionally, AuTx® and Kevlar KM2 exhibited a high degree of fibrillation when impacted with both the FSP and round projectile heads, and Kevlar KM2 demonstrated a correlation between impact velocity and rupture surface geometry. At low velocities, KM2 demonstrated a highly fibrillated rupture surface, quite similar to impact with the round projectile, while higher impact velocities yielded a trend towards a more localized failure, being somewhat similar to impact with the razor blade projectiles. Although Dyneema® SK76 exhibited cutting when impacted with the razor blade projectiles, impact with both FSP and round projectiles exposed a high-degree of melting at all impact velocities, either due to local heating during contact or failure, or possibly due to post-failure snap-back. Finally, it is suggested that findings from the current experimental data
set can ultimately be used to aid in understanding of local yarn failure during full fabric impact with an emphasis placed on performing quasi-static transverse loading experiments or dynamic critical velocity experiments on candidate filament material during the initial research and qualification phase of novel fiber development.
7. Exploration of Wave Development During Yarn Transverse Impact

Adapted from:

7.1 Abstract

Single yarns have been impacted in a transverse fashion so as to probe the characteristics of resulting wave development. Longitudinal wave speed is tracked in efforts to directly measure the yarn tensile stiffness, resulting in a slight increase in stiffness of Kevlar® KM2 and Dyneema® SK76. Additionally, the load developed in AuTx® and Kevlar® KM2 yarns behind the longitudinal wave front has been tracked, providing additional verification for the Smith relations. Further effort to bolster the Smith equations has performed by tracking transverse wave speeds in AuTx® yarns, which agree very well with the Smith equations for a wide range of impacting velocities. Additional emphasis has been placed at understanding the transverse wave development around the yarn critical velocity, demonstrating that there is a velocity zone where partial yarn failure detected. Above the critical velocity, measurement of early time transverse wave speeds also agree with the Smith solution, though the wave speed quickly reduces in value due to the drop in tensile stresses resulting from filament rupture. Finally, the Smith equations have been simplified and are compared to the Cunniff equation, which bear a striking resemblance. Due to such resemblance, it is suggested that yarn critical velocity experiments can be performed on trial yarn material and the effect of modifying yarn mechanical properties is discussed.
7.2 Introduction

Although a rather uncommon experiment, transverse impact into single yarns has been historically used to determine baseline yarn mechanical properties, specifically to determine yarn stiffness in the longitudinal direction and the yarn critical velocity, wherein the material fails "instantly" upon projectile-yarn contact. Specifically, such an understanding of stiffness and critical velocity is used to assess the efficacy of implementing a specific yarn material into a full body armor system. Such an efficacy analysis is most reasonably performed by assessing the Cunniff parameter \((\Omega^{1/3})\) [10], which is described in Equation 7.17,

\[
\Omega^{1/3} = \left(\frac{\sigma \epsilon}{2\rho} \sqrt{\frac{E}{\rho}}\right)^{1/3}
\]

wherein \(\sigma\), \(\epsilon\), \(E\), and \(\rho\) represent the longitudinal failure strength, longitudinal failure strain, longitudinal elastic modulus, and density, respectively. The Cunniff parameter can be best thought of as the product of the yarn specific toughness and the longitudinal wave speed; essentially, it is desired to maximize the energy absorbed by the constituent yarns before rupture, and to move this energy away from the impact site as quickly as possible. Thus, maximizing strength, strain, and stiffness, while decreasing density, are key parameters required to increase the halting capability of a fabric system. Further analysis into the coupling of the first three parameters is discussed below in Section 7.5.

Understanding the Cunniff parameter then leads back to the power available in performing single yarn impact experiments. Not only are yarn experiments much cheaper to perform on novel materials as compared to full fabric experiments, it is actually possible to determine the failure strength, failure strain, and elastic modulus developed during the single-yarn impact event. Each of the aforementioned mechanical properties can be directly input into the Cunniff parameter, thereby allowing comparison of various yarn types in an environment creating similar loading conditions to that seen from full fabric impact. It must be noted that traditional pre-
dictions of the Cunniff parameter generally use quasi-static mechanical properties, but as was shown by Hudspeth et al. [160], care must be taken to understand the projectile loading conditions exhibited on the yarn of interest.

Having provided explanation for need to ascertain input values for the Cunniff parameter described in Equation 7.17, it is now of interest to briefly overview the governing mechanics of transverse impact of a projectile into a single yarn in efforts to provide a background on the deformation behavior and resulting system response. Upon impact, a longitudinal tensile wave $c$ emanates away from the projectile-yarn contact site, moving at the longitudinal speed of sound in the material, being described by Equation 7.2, where $E$ and $\rho$ represent the longitudinal tensile modulus and the material density, respectively.

$$c = \sqrt{\frac{E}{\rho}}$$  \hspace{1cm} (7.2)

Material behind the longitudinal wave front is set in tension, moving inwards towards the projectile-yarn contact site with a particle velocity of $W$, which is described by Equation 7.3, wherein $\epsilon$ signifies the strain amplitude developed due the passing of the longitudinal wave.

$$W = c\epsilon$$  \hspace{1cm} (7.3)

An additional wave is also developed upon the projectile-yarn contact, following behind the longitudinal wave, and possesses the velocity $U$, being described by Equation 7.4, or in the laboratory reference frame, as $U_{obs}$, being described by Equation 7.5.

$$U = \sqrt{\frac{E}{\rho}} \left( \frac{\epsilon}{1 + \epsilon} \right)^{1/2}$$  \hspace{1cm} (7.4)

$$U_{obs} = c(\sqrt{\epsilon(1 + \epsilon)} - \epsilon)$$  \hspace{1cm} (7.5)
Material behind the transverse wave front transitions from an inward velocity (perpendicular to the projectile direction) to a particle velocity identical in magnitude and direction to that of the projectile, being described by Equation 7.6. Solving Equations 7.2-7.6 allows for determination of the strain developed in the yarn at a specific impacting velocity, \( V \), or vice-versa, one can solve for the velocity required to initiate a specific strain value, \( \epsilon \).

\[
V = \left( (1 + \epsilon)^2 U^2 - (1 + \epsilon)U - W \right)^{1/2}
\]

Finally, Equation 7.7 can be used to solve for the angle \( \theta \) developed behind the transverse wave front, which notably remains constant during the impact event, barring reverberations of waveforms from the boundaries at the yarn ends.

\[
\theta = \tan^{-1} \left( \frac{V}{U(1 + \epsilon) - W} \right)
\]

The validity of Equations 7.2-7.7 are reasonably well substantiated by experimental evidence from various authors. The reader is directed to Table 6.1 of Chapter 6 for a thorough list of previous authors who used the single yarn transverse impact experiment to gain understanding of yarn mechanical properties, specifically the yarn critical velocity. Although solely a selection from the data set listed in Table 6.1, the reader is directed to work performed by Chocron et al. [53] and Chocron and Walker [54], who take great effort in determining the critical velocity for several types of high-performance yarn. Of importance to note are the efforts taken by Walker and Chocron [54] to understand the yarn critical velocity, wherein they present a strong explanation for the difference between experimental and theoretical critical velocities predicted by Equations 7.2-7.2. Further analysis of the critical velocity is has also been performed by Carr [50], Bazhenov et al. [52], Sockalingam et al. [143], and Hudspeth et al. [160].

The current experimental data set was analyzed in efforts to determine the presence of a rate sensitivity for both Kevlar\textsuperscript{R} KM2 and Dyneema\textsuperscript{R} SK76, but concurrent
efforts also led to experiments analyzing the validity of the strain state developed behind the longitudinal wave front along with the velocity of the transverse wave, with both sets of latter experiments being performed on AuTx® yarn. Finally, due to the nature of determining the in-situ mechanical properties of these materials (along with data from Hudspeth et al. [2016]), it is of interest to assess the effect of varying stress, strain, and modulus in efforts to achieve greater transverse critical velocities, which will be discussed in Section 7.5. As previously mentioned, this variation of mechanical properties has also been expanded to briefly assess their effect on the oft-quoted Cunniff parameter.

7.3 Experimental

7.3.1 Materials

Three different high performance yarn materials have been used to assess both the longitudinal and transverse stress wave characteristics, namely Dyneema® SK76, Kevlar® KM2, and AuTx®. Further explanation of these materials and reasoning for their usage in the current experimental set can be found in Section 6.3.1 of Chapter 6. Yarns have been impacted with three different projectile nose geometries, namely razor blade, FSP, and round. An image of the three different projectile heads can be found in Figure 6.3 of Chapter 6.

7.3.2 Longitudinal Wave

In order to determine the presence of a rate sensitivity of high-performance fiber in a transverse impact environment, the velocity of the longitudinal wave fronts developed upon projectile-yarn contact wave were tracked in-situ. Initially, it is of course valid to assess the yarn wave speed, \( c \), via determining the time required for the longitudinal wave to travel from the contact site to the yarn grip, which has been mounted onto a fast-response force transducer. This simple method is described by Equation
7.8, where $L$ represent the distance from the projectile contact site to the yarn clamp and $t$ represents the time of impact to the time of arrival of the longitudinal wave to the yarn clamp.

$$c = \frac{L}{t} \quad (7.8)$$

As described in Chapter 6, high speed imaging has been used to track yarn deformation and failure during impact event, which ideally, could give an impact time, but it must be noted that it is impossible to know the actual time of impact of the projectile into the yarn, as contact will always be made at some time between frames or during a frame exposure. Thus, while estimating the contact time using imaging would be appropriate with sufficient framing rates, low frame rates result in inter-frame separation great enough to shadow the initial contact time, resulting in large bounds being placed on the measured wave velocity. As such, estimating the time of impact has been avoided in lieu of an alternative measurement approach, namely using the arrival times of the longitudinal wave fronts at two ends of the yarn that have a large offset between the yarn mid-length and the actual contact site. An example of this offset impact geometry can be seen in Figure 7.1. In this specific geometry it is possible to accurately determine the location of impact with a reasonably high level of accuracy using a laser bore site inserted into the muzzle of the barrel. Measuring from the clamp locations to the location demarcated by the laser yields both lengths $L_1$ and $L_2$. The wave velocity $c$ can then be determined using Equation 7.9 if one were to know the time from impact to the arrival of the stress wave at the clamp boundary, being denoted as $t_1$ and $t_2$, respectively.

$$c = \frac{L_1}{t_1} = \frac{L_2}{t_2} \quad (7.9)$$

As previously stated, without sufficient framing rates, both $t_1$ and $t_2$ can only be estimated with a low level of precision. In order to circumvent this high variability measurement, a more appropriate method has been pursued wherein one tracks the
Figure 7.1. Schematic of single yarn transverse geometry used to determine longitudinal wave speeds.
difference in longitudinal wave travel times from projectile-yarn contact site to detection of the longitudinal wave from the clamping load cells for drastically different yarn lengths $L_1$ and $L_2$. Such a measurement is described in Equation 7.10 and is demonstrated in Figure 7.2. As previously stated, possessing accurate $L_1$ and $L_2$ is achieved using a laser bore site, allowing for Equation 7.10 to provide an accurate estimate the wave speed of the yarn material. As an aside, it is important to note that the longitudinal stress wave must also travel through the yarn clamping fixtures, but accounting for this travel time is unnecessary if identical clamping fixtures are used on both ends of the yarn; the longitudinal wave travel time through the clamps is identical in both fixtures and is inherently subtracted out using aforementioned time difference method.

$$c = \frac{L_2 - L_1}{\Delta t}$$  \hspace{1cm} (7.10)

Finally, due to the presence of the aforementioned load cells onto which the yarn clamps were attached, it was feasible to track the load developed in the yarn behind
the longitudinal wavefront. Specifically, effort was placed on determining the stress developed in the yarn as a comparison to the analytical evaluation of strain described in Equations 7.2. Results from razor blade transverse impact was analyzed for both Kevlar® KM2 and AuTx® yarns. It is important to note that although the maximum load detected by the transducers does indeed represent the max load generated by the longitudinal stress wave, the temporal load evolution depicted by the force transducers is not identical to that present in the actual fiber, due to the inertial presence of the yarn clamp; caution is advised when analyzing the load history in Figure 7.2, as a reader may initially believe it is possible to assess a strain rate from the slope of the force history plot. The quantity of interest from this data set is based on the max load amplitude, and for example, for Figure 7.2 is roughly 44 N. Additionally, it is important to note that the force values measured in the load cells are \( \sim \) twice that generated in the wake of the longitudinal wave, due to the wave reflection from the rigid clamp boundary. As such, the measured load values used for analysis of Equation 7.2 result from maximum detected load amplitudes that have been appropriately halved to represent the load that would be developed behind the longitudinal wave front before interaction with the rigid clamping boundary.

### 7.3.3 Transverse Wave

Along with determining the effect of impact velocity on the longitudinal wave speed, it was also determined of use to assess the transverse wave velocity as a function of impact velocity. This transverse wave, which is described by Equation 7.4 in Section 7.2, presents itself in a tent formation, thereby allowing direct tracking of the wave velocity using high-speed imaging. An example of this transverse wave movement can be seen in the schematic shown in Figure 7.3. Specifically, attention was placed on movement of \( U_{\text{location}} \) to \( U_{\text{location}}^* \) between several frames, and in tandem with known inter-frame separations, observed transverse wave velocities were calculated for various impacting velocities.
Figure 7.3. Longitudinal and transverse wave front positioning at two times post impact.
7.4 Results and Discussion

7.4.1 Longitudinal Wave

Wave Speed

As described in Section 7.3.2, longitudinal wave speeds have been tracked for a number of transverse impact experiments for Dyneema® SK76 and Kevlar® KM2. A table describing wave speeds and corresponding elastic moduli can be found in Table 7.1.

Transverse impact experiments performed on Dyneema® SK76 yarn resulted in longitudinal wave speed values of 11466.34±794.71 m/s, 12416.00±342.44 m/s, and 12331.57±261.61 m/s, when impacting with razor blade, FSP, and round projectile geometries, respectively. Although small, it is suggested that the variation in wave speed could arise for two reasons. First, the projectile contact site may vary slightly, with one standard deviation in the measured wave speed resulting in a contact site variation of ∼2-7 mm. Though plausible, such a large shot line deviation is not seen in any of the high-speed images. An alternate and more realistic explanation is that the material is slightly rate sensitive, resulting in an increase in longitudinal wave speeds at higher impacting velocities, being demonstrated in Figure 7.4(a). Such rate sensitivity is not uncommon in UHMWPE fiber, and a good example is given by Cansfield et al. [86] who showed a linear increase in failure stress with a logarithmic increase in applied strain rate from strain rates ranging from $10^{-4}$ s$^{-1}$ to $10^{-1}$ s$^{-1}$.

Finally, it is important to note that the elastic modulus exhibited by these fibers at such an elevated strain rate has been determined to be roughly 128 GPa, 150 GPa, and 148 GPa when using the razor blade, FSP, and round indenters. Again, similar to the measured variation in wave speed, the demonstrated increase in elastic modulus most likely occurs due to the increase in impacting velocity, thereby increasing the strain rate at which the filaments are loaded in tension. It is highly doubtful that the projectile nose geometry plays any role on the longitudinal wave speed, rather,
Table 7.1. Wave speeds and resulting moduli for both Dyneema® SK76 and Kevlar® KM2.

<table>
<thead>
<tr>
<th>Fiber Type</th>
<th>Kevlar® KM2</th>
<th>Dyneema® SK76</th>
</tr>
</thead>
<tbody>
<tr>
<td>Density (kg/m³)</td>
<td>1440</td>
<td>970</td>
</tr>
<tr>
<td>Projectile</td>
<td>Razor Blade</td>
<td>FSP</td>
</tr>
<tr>
<td>Wave Speed - Avg (m/s)</td>
<td>8181.53</td>
<td>8454.03</td>
</tr>
<tr>
<td>Wave Speed - Std (m/s)</td>
<td>127.94</td>
<td>92.61</td>
</tr>
<tr>
<td>Elastic Mod - Avg (GPa)</td>
<td>96.39</td>
<td>102.92</td>
</tr>
<tr>
<td>Elastic Mod - Std Low (GPa)</td>
<td>93.40</td>
<td>100.68</td>
</tr>
<tr>
<td>Elastic Mod - Std High (GPa)</td>
<td>99.43</td>
<td>105.18</td>
</tr>
</tbody>
</table>

The three different geometries used for the current experiments were shot generally in different velocity regimes, as these experiments were primarily performed in order to determine the yarn critical velocity, as described in Chapter 6.

Wave speeds for Kevlar® resulted in values of $8181.53 \pm 127.94$ m/s, $8454.03 \pm 92.61$ m/s, and $8505.15 \pm 53.30$ m/s, for the razor blade, FSP, and round projectile geometries, respectively. Similar to the explanation given for impact into Dyneema® yarns, it is suggested that the variation in wave speed may have arisen due to a slight projectile contact site variation, with one standard deviation in the measured wave speed resulting in a variation of $\sim 1$ mm, but the presence of a rate sensitivity is much more plausible. This slight degree of material rate sensitivity is demonstrated by the increase in longitudinal wave speeds at higher impacting velocities, which is demonstrated in Figure 7.4(b). Using the experimentally measured wave speeds exhibited by these fibers at such an elevated strain rate, calculation of the elastic modulus results in values of roughly 96 GPa, 101 GPa, and 103 GPa when using the razor blade, FSP, and round indenters, respectively. Such an increase in modulus, as just mentioned, most likely occurs due to the increase in impacting velocity ranges of each projectile, thereby increasing the strain rate at which the filaments are loaded in tension.

**Longitudinal Stress**

In addition to determining the velocity of the longitudinal wave, it was also deemed of importance to track the stress developed behind the longitudinal wave front. Specif-
Figure 7.4. Measured wave speeds of (a) Dyneema® and (b) Kevlar® KM2.
ically, it was of desire to probe the validity of Equations 7.2-7.6. As such, due to the use of force transducers required for camera triggering and longitudinal wave tracking, it became possible to track the force developed behind the longitudinal wave front when impacting at low velocities. This low velocity stipulation, being required by load cell restrictions, led to tracking of force levels developed due to transverse impact of razor blade projectiles, as such sharp projectiles presented the lowest yarn critical velocity transition regime [160]. Two different yarn types were analyzed when impacting in the aforementioned loading conditions, namely Kevlar® KM2 and AuTx®, with resulting stress histories shown in Figure 7.5. Additionally, the analytical stress described by Equations 7.2-7.6 has also been overlaid on both plots within Figure 7.5. It is important to note that the stress measured in the load cell is \( \sim \) twice that generated in the wake of the longitudinal wave, due to the wave reflection from the rigid clamp boundary. As such, the measured stress values shown in Figure 7.5 result from maximum detected load amplitudes that have been appropriately halved to represent the stress that would be developed behind the longitudinal wave front before interaction with the rigid clamping boundary. To the author’s knowledge, this is the first instance wherein direct measurement of the load behind the longitudinal wave front has been performed.

7.4.2 Transverse Wave Speed

In order to provide additional validity of Equations 7.2-7.6, it was deemed of use to track the transverse wave speed developed in yarn when subjected to transverse impact. As such, transverse wave speeds were analyzed via high-speed imaging of impact into AuTx® yarns using both FSP and round projectile heads. Implemented impact velocities spanned from roughly 400 m/s to 1000 m/s. Below the critical velocity, wherein no filaments fail upon impact, impact via FSP and round projectile heads resulted in transverse wave speeds being quite close to those predicted from Equation 7.5. Examples of this impact event can be seen in Figures 7.6 and 7.7, when
Figure 7.5. Longitudinal tensile stress developed behind the longitudinal wave front in both (a) Kevkar® KM2 and (b) AuTx®.
Figure 7.6. (a) Image sequence of a FSP transversely impacting into a single AuTx® yarn at 394 m/s (below critical velocity transition regime [160]), along with (b) corresponding measurement of transverse wave speed overlaid with the theoretical value predicted by Equation 7.5.

impacting with FSP and round projectiles, respectively. Both image sequences, which represent impacting velocities of 394 m/s (FSP) and 508 m/s (Round), demonstrate experimental transverse wave velocities quite close to their respective theoretical values, being 824 m/s compared to 832 m/s (FSP) and 960 m/s compared to 970 m/s (Round).

At higher impacting velocities residing within the critical velocity transition region, the lead transverse wave velocity again showed very good agreement to theo-
Figure 7.7. (a) Image sequence of a round projectile transversely impacting into a single AuTx® yarn at 508 m/s (below critical velocity transition regime [160]), along with (b) corresponding measurement of transverse wave speed overlaid with the theoretical value predicted by Equation 7.5.
Theoretical prediction, which can be seen in Figures 7.8 and 7.9 for 621 m/s FSP and 659 m/s round projectile impact, respectively. The measured experimental and theoretical transverse wave velocities were 1142 m/s and 1093 m/s for the FSP and 1114 m/s and 1132 m/s for the round projectile, respectively. Interestingly, both projectiles clearly demonstrate the partial nature of failure when shooting within the critical velocity transition region [160], and a range of transverse wave velocities are detected, depending on the fail/no-fail status of each filament. Both Figures 7.8 and 7.9 also show a plot of the minimum measured transverse wave speed, thereby demonstrating the aforementioned variation in transverse wave speed. Said minimums have been measured as 223 m/s and 239 m/s for the FSP and round projectile impact conditions shown in Figures 7.8 and 7.9, respectively. It is important to note that the minimum and maximum measured wave speeds appear to be constant throughout the range of relevant image times.

Transverse wave speeds were also tracked from representative shots fired above the critical velocity transition regime for both FSP and round projectile geometries, which is shown in Figures 7.10 and 7.11. As opposed to the shots fired below and within the critical velocity transition regime, shots impacting above the critical velocity region do not express constant transverse wave velocities. In contrast, they show a progressive reduction in transverse wave speed with increasing time post projectile-yarn contact. Interestingly, for both the FSP and round projectile impact, the initial measured transverse wave velocity appears to correspond exceedingly well with the theoretical transverse wave velocities predicted from Equation 7.5. Initial measured transverse wave speeds and the theoretical wave speeds were 1281 m/s and 1273 m/s for the FSP impact and 1316 m/s and 1433 m/s for the round projectile impact, respectively. The FSP impact velocity was measured to be 806 m/s and the round impact velocity was measured to be 994 m/s. For both impact conditions, a stark drop in transverse wave speed can be seen in the measured time window, resulting in a transverse wave velocity of 748 m/s 12.2 μs from impact and 749 m/s 11.8 μs from impact, for the FSP and round projectile respectively. As with the shots...
Figure 7.8. (a) Image sequence of a FSP transversely impacting into a single AuTx® yarn at 621 m/s (inside critical velocity transition regime [160]), along with (b) corresponding measurement of transverse wave speed overlaid with the theoretical value predicted by Equation 7.5.

shown in Figures 7.8 and 7.9, there exists a range in the transverse wave speed due to partial yarn failure. The slowest moving broken filaments appear to possess a transverse wave speed of roughly 200-400 m/s, with this wave speed reducing in value as time post contact progresses. Again, it is reiterated that there exists a reduction in both maximum and minimum transverse wave speeds with progressing time post impact. Such a reduction in wave speed values is believed to signify the occurrence
Figure 7.9. (a) Image sequence of a round projectile transversely impacting into a single AuTx® yarn at 659 m/s (inside critical velocity transition regime [160]), along with (b) corresponding measurement of transverse wave speed overlaid with the theoretical value predicted by Equation 7.5.
Figure 7.10. (a) Image sequence of a FSP transversely impacting into a single AuTx yarn at 806 m/s (above critical velocity transition regime [160]), along with (b) corresponding measurement of transverse wave speed overlaid with the theoretical value predicted by Equation 7.5.

of progressive failure in the yarn; indeed, initial transverse wave speeds are quite similar to the corresponding theoretical transverse wave speed, demonstrating intact filaments, which is then followed by complete filament failure leading to a reduction in demonstrated transverse wave speed.

Finally, transverse wave speeds measured from several experiments within the 400 m/s to 1000 m/s velocity window were plotted against their corresponding impact velocity, and can be found in Figure 7.12. As shown in the data presented in Figures 7.6 through 7.11, the transverse wave propagation measurement scheme was slightly different below, within, and above the critical velocity transition regime. Below said regime, the transverse wave front was tracked as the entire yarn showed a stark "tent-like" deformation and stayed constant during the entire imaging history. Within the
Figure 7.11. (a) Image sequence of a round projectile transversely impacting into a single AuTx\textsuperscript{®} yarn at 994 m/s (above critical velocity transition regime \[160\]), along with (b) corresponding measurement of transverse wave speed overlaid with the theoretical value predicted by Equation 7.5.
transition regime a range of critical velocities were measured due to partial yarn failure, and only the leading edge of the transverse wave front was used to calculate the wave speed presented in Figure 7.12. Above the transition regime, due to rapid failure of the entirety of the yarn, only early time calculations were used to determine the transverse wave velocities presented in Figure 7.12. Overlaid on top of the experimental transverse wave velocities in Figure 7.12 is also the theoretical transverse wave velocity as a function of the projectile impact speed. The experimentally measured velocities correspond quite well with the theoretical transverse wave speed, bringing further verification to Equations 7.2 - 7.6.

### 7.5 Optimization of Filament Properties

Due to the conjectures made regarding the variation in mechanical properties of high-performance filament material, and the corresponding result on the performance
of a fabric system, a basic thought experiment has been performed to assess the
effect of varying failure strength, failure strain, and elastic modulus of yarn material,
both in controlling the resulting critical velocity as predicted by the Smith equations
(Equations 7.2-7.7) [20] and even more importantly, the Cunniff parameter (7.17).

7.5.1 Constitutive Equation Modification

Due to the simultaneous solution requirement of Equations 7.2-7.6, said relations
have been somewhat simplified via removing higher order strain terms. Such a sim-
plification can be seen in the following:

\[ c = \sqrt{\frac{E}{\rho}} \quad (7.11) \]

\[ W = C\epsilon \quad (7.12) \]

\[ U = \sqrt{\frac{E}{\rho}} \left( \frac{\epsilon}{1 + \epsilon} \right)^{1/2} \cong c\sqrt{\epsilon} \quad (7.13) \]

\[ U_{obs} = c(\sqrt{\epsilon(1 + \epsilon)} - \epsilon) \cong c(\sqrt{\epsilon} - \epsilon) \quad (7.14) \]

\[ V = ((1 + \epsilon)^2U^2 - ((1 + \epsilon)U - W)^2)^{1/2} \cong c\sqrt{2\epsilon\sqrt{\epsilon}} \quad (7.15) \]

Although not a true linearization of Equations 7.4-7.6, such a simplification does
indeed reduce the effort needed to arrive upon a solution, as a numerical solution
procedure is not needed to solve for \( \epsilon \) given \( V \). In order to demonstrate the validity
of Equations 7.13-7.15, all three relations have been solved for a range of impact
velocities up to 1 km/s and are plotted in Figure 7.13. Figure 7.13(a) compares
the full solution (7.6) and simplified solution (7.15) of the tensile strain developed
behind the longitudinal wave front, showing very good agreement, with the simplified
solution tending to be slightly less than the full solution. Figure 7.13(b) compares the variation in the full solution and simplified solution of the transverse wave speed in both a reference onboard the yarn (Equation 7.4 to Equation 7.13, respectively) and from the reference of a laboratory observer (Equation 7.5 to Equation 7.14, respectively). Both comparisons show very similar results between the full solution and simplified solution with the partial solution $U$ tending to slightly over predict the full solution, and the partial solution of $U_{obs}$ tending to slightly under prediction the full solution. Thus, the following assessment of mechanical property evaluation on the resulting critical velocity uses the simplified set of equations for ease of analysis.

### 7.5.2 Material Property Modification - Single Yarn Critical Velocity

As it has been proposed that an increase in yarn critical velocity will result in a corresponding increase in full fabric performance, it is of interest to understand the coupled effect of increasing single filament mechanical properties. Specifically, it is of interest to understand the effect of varying failure stress, failure strain, and elastic modulus. As a first trial of single filament performance, it is of use to compare linear elastic mechanical parameters, considering the long-range tensile characteristics demonstrated at rupture. Although this analysis assumes a linear elastic response of the material, such an assumption is well substantiated from experimental evidence [35,36,89]. It must be also be noted that previous work has demonstratively shown single filaments do not fail in pure tension [130,139], but this short analysis is solely being performed to elucidate the effect of local failure characteristics resulting in strength/strain increases measured during transverse loading. Specifically, aim is directed at the statement made from Phoenix and Porwal for the need to decrease the yarn longitudinal wave speed while increasing yarn strength in order to increase the Cunniff parameter $\Omega^{1/3}$.

In order to assess effect of the coupled variation of failure strength, $\sigma$, and failure strain, $\epsilon$, on the resulting yarn critical velocity, Equation 7.15 has been modified to
Figure 7.13. Plots showing the variation in the full analytical solution and simplified solution of (a) the longitudinal strain and (b) the transverse wave speed developed during transverse impact. Material properties for the given plots represent a yarn material possessing an elastic modulus of 130 GPa and a density of 1450 kg/m$^3$. 
remove the elastic modulus term, \( E \), present within the yarn wave speed \( c \), thereby presenting the critical velocity, \( V \), as a function of solely \( \sigma \) and \( \epsilon \). Said modification can be seen in Equation 7.16.

\[
V = \sqrt{\frac{2}{\rho} \sigma \sqrt{\epsilon}}
\]  

(7.16)

An array of values of \( \sigma \) and \( \epsilon \) are then input into Equation 7.16, yielding a critical velocity surface, which can be seen in Figure 7.14(a). Additionally, iso-velocity curves have been overlaid on top of the velocity surface with curves ranging from 300 m/s to 1000 m/s in 100 m/s increments. Take notice of the clear affect which both failure strain and failure stress have on the resulting critical velocity. Increasing either value while holding the other constant shows an increase in critical velocity, but simultaneous increases in both are clearly the most preferable.

In efforts to more easily visualize the effect of both failure stress and failure strain, Figure 7.14(a) has been reoriented to view along with the critical velocity axis, which can be seen in Figure 7.14(b), using the same color scheme present in Figure 7.14(a). As previously stated, there is clearly a coupled effect of increasing either the long-range failure stress or failure strain of the impacted yarn material, assuming a linear elastic response. The described plot contains iso-velocity lines for a filament material possessing a density of 1.45 g/cm\(^3\). Overlain on said plot can be seen an example failure stress/strain solution of a yarn exhibiting a critical velocity of 500 m/s. In this specific case, the failure stress and failure strain are \( \sim 1.48 \) GPa and 1.48%, respectively. Again, note that an increase in either failure property (failure stress or failure strain) while holding the neighboring parameter constant, will result in an increase the transverse impact critical velocity.

It is also interesting to mention that changes in failure stress result in a more rapid increase (and decrease) in critical velocity, as compared to proportional changes in failure strain. For example, a 10% increase in failure stress results in a critical velocity of 524.4 m/s, while a 10% increase in failure strain results in a critical velocity of 512.1
Figure 7.14. Variation in the critical velocity of a single yarn subject to transverse impact ($V_{\text{critical}}$) when varying both the failure stress and failure strain of the AuTx® yarn material.
m/s. Figure 7.15 demonstrates the variation in critical velocity due to modification of either exclusively failure stress or exclusively failure strain.

### 7.5.3 Material Property Modification - Cunniff Equation

Similar to the analysis performed on the simplified Smith equations given in the previous section, it was also deemed of use to the analyze the effect of changes in mechanical properties on the resulting Cunniff parameter, which is shown in Equation 7.17.

\[
\Omega^{1/3} = \left(\frac{\sigma}{2\rho}\sqrt{\frac{E}{\rho}}\right)^{1/3}
\]  

Equation 7.17 can also be modified to be written as a coupled function of either stress or strain and the longitudinal wave speed. As previously mentioned, such a
form was used by Phoenix and Porwal [135] in order to assess the effect that changing mechanical properties has on the resulting Cunniff parameter.

\[
\Omega^{1/3} = \left( \frac{\epsilon}{\sqrt{2}} \right)^{2/3} c = \left( \frac{\sigma}{\sqrt{2} \rho} \right)^{2/3} c^{-1/3}
\]

(7.18)

The first inequality demonstrates the effect of varying either failure strain or elastic modulus. In this format, increases in ballistic performance are initially suggested to increase when their is either an increase in failure strain or longitudinal wave speed, which follows intuitive reason. In contrast, the latter equality suggests that either increases in failure stress or decreases in longitudinal wave speed will render increases in the resulting Cunniff parameter. Although formally correct, such a representation of the Cunniff parameter is somewhat misleading and reasoning for odd occurrence is the result of the failure stress having a greater effect on the Cunniff parameter as opposed to the failure strain, which will be demonstrated shortly. First though, similar to the data set shown for the Smith equation, it is of use to plot out the Cunniff parameter as a function of both failure stress and failure strain, which can be seen in Figure 7.16(a). Additionally, iso-velocity lines have been overlaid on the surface plot of Figure 7.16(a), ranging from 100 m/s to 700 m/s in 100 m/s intervals. In efforts to more easily visualize the effect of failure stress and failure strain on the resulting Cunniff parameter, the 3D surface plot of Figure 7.16(a) has reoriented to a 2D color map showing, which is depicted in Figure 7.16(b), Iso-velocity lines similar to those depicted in Figure 7.16(a) have also been overlaid on Figure 7.16(b).

Now it is of interest to assess the previously mentioned statements made by Phoenix and Porwal [135], which originate from Equation 7.18. First, a linear elastic stress-strain curve having a modulus of 100 GPa has been overlaid on Figure 7.16(b), which shows a Cunniff parameter of 400 m/s, resulting in a failure stress and failure strain of 1.49 GPa and 1.49%, respectively. It must be pointed out that an increase in either failure stress or failure strain will provide an increase in ballistic performance, but increases failure stress demonstrate a greater effect than proportional changes in failure strain; this effect can be seen in Figure 7.17, wherein a 10% increase in solely
Figure 7.16. Variation in the Cunniff Parameter, $\Omega^{1/3}$, when varying both the failure stress and failure strain of the constituent AuTx® yarn material.
failure strain or a 10% increase in solely failure stress exhibits a Cunniff parameter increase of 1.6% and 4.9%, respectively. Herein lies the reasoning for the statement made by Pheonix and Porwal, wherein it is suggested that a filament material modified to exhibit a greater failure stress and a reduction in elastic modulus would be of benefit. Although valid, such the author suggests that such a statement can easily lead an inexperienced reader astray. A fiber manufacturer should not want to create a filament that reduces wave speed. Rather, he/she should want to increase filament toughness and the filament wave speed. Holding the failure stress constant and decreasing the wave speed is an odd way to describe an increase in toughness, coming from an increase in failure strain. Inherently in this procedure, the wave speed is decreased, but not because this is desired, but rather because this is a material trade-off. We have increased toughness, which has more control on the Cunniff parameter, via sacrificially decreasing the wave speed, which does not contain as much control on the Cunniff parameter. The greatest concern from the present author is the statement, “increasing the strength per unit weight of the fiber material is crucial, and if at the same time the tensile wave speed can be decreased, so much the better.” Although valid for a perfectly linear elastic material, it seems much more useful state something more along the lines, ”increasing the strength per unit weight of the fiber material is crucial, and if at the same time the failure strain be held constant or even increased, so much the better,” as having a decrease in modulus with an increase failure stress is really the same as having a simultaneous increase in failure stress and failure strain. Finally, this statement made by Pheonix and Porwal considers a material that is purely linear elastic. New UHMWPE material have been shown to actually exhibit a convex stress-strain response, and as such, the aforementioned statement then becomes quite invalid. It would be preferable to have material that demonstrates an unchanged failure stress and failure strain, with an increase modulus. Although purely personal preference, the current author feels it is much more appropriate to push readers to analyze filament properties from the standpoint of the original Cunniff parameter, namely the product of specific toughness and wave
speed (Equation 7.17) as opposed to modified forms (Equation 7.18) that may lead the reader astray.

Finally, it is of use to compare both the Smith equation and the Cunniff equation in the format shown in Equations 7.19 and 7.20.

\[
V = \sqrt{\frac{2}{\rho}} \sigma^{1/2} \epsilon^{1/4} \quad (7.19)
\]

\[
U^{1/3} = \sqrt{\frac{2^{-1/3}}{\rho}} \sigma^{1/2} \epsilon^{1/6} \quad (7.20)
\]

Interestingly both equations are very similar in nature, with the Cunniff equation being slightly less controlled by the failure strain than the Smith equation. The similarity of these equations can also be verified by the plots shown in Figures 7.14 and 7.16. It is thus suggested that very similar understanding can be gained from finding the yarn critical velocity by performing single yarn impact experiments as
compared to shooting entire armor systems in efforts to determine $V_{50}$ values. Used as a preliminary design tool, the former experiment is much less costly to perform, and analysis can be performed on small batches of material developed either from desktop made filaments, or from pilot runs on an experimental fiber spinning line.

7.6 Conclusions

Initial portions of this work have been directed at performing single yarn transverse impact experiments in efforts to gain understanding of waves developed due to the impact event. Firstly, longitudinal wave speeds were tracked in Kevlar® KM2 and Dyneema® SK76 in order to back-out material stiffness, ultimately with the desire to detect the presence of rate sensitivity. Three different projectile nose geometries have been used to impact the yarn material, namely razor blade, FSP, and round. Differing wave speeds were detected from the three different projectile nose geometries, not due of the nose geometry itself, but rather because of the differing strike velocities used for each nose geometry. As these experiments were performed in efforts to uncover the yarn critical velocity [160], increasing impact velocities were required to promote immediate rupture for Razor blade, FSP, and round nose geometries, respectively, due to the corresponding increase in stress concentration for the three projectile heads. Longitudinal wave speed measurements were also used to detect the max load developed in Kevlar® KM2 and AuTx® yarn behind the longitudinal wave front, which agreed very well theory. Subsequently, transverse wave speeds were measured for AuTx® yarns for a variety of impacting velocities, which too agreed quite well with theory. Effort was also placed at understanding the transverse wave development around the yarn critical velocity. Below the critical velocity, the transverse wave demonstrated full yarn movement with a constant transverse wave speed. Within the critical velocity transition region, partial yarn failure was detected, resulting in a range of transverse wave speeds exhibited by the material; the leading edge of the transverse wave wave was constant and moved with the theoretically predicted velocity. The
slowest moving transverse wave, though constant in the region measured, was much slower than the leading edge, presumably due to the lack in tension because of filament failure. Above the critical velocity, the entire yarn failed, but measurement of the initial leading edge of the transverse wave yielded a transverse wave velocity similar to that predicted from theory. With time progression, the leading edge (and the trailing edge) of the transverse wave decreased in velocity.

Finally, effort has been directed towards understanding room for filament improvement, demonstrating that linear elastic fibers promote greater increases in ballistic performance (both Cunniff parameter and the single yarn critical velocity) with increases in yarn toughness or elastic stiffness, although the former yields greater performance effects than the latter. Additionally, it was determined that changes in failure stress are more significant than changes in failure strain. Due to the similarity between the Cunniff parameter and the critical velocity arising from the Smith equations, effort has also been directed towards warning the reader to be wary of manipulating the Cunniff equation in a way to show that decreases in wave speed can actually improve ballistic performance. Although correct in certain modifications of failure stress, such a claim actually is just suggesting an increase in yarn toughness, which as previously mentioned, is more powerful at controlling the Cunniff parameter than the elastic stiffness.
8. Local Deformation and Failure of Yarn Subjected to Transverse Impact


8.1 Abstract

Single Kevlar® KM2 and Dyneema textsuperscript® SK76 have been impacted with both flat 20 mm projectiles and round 20 mm projectiles. As opposed to all known previous yarn transverse impact studies, specific attention has been placed on varying the viewing angle of the camera system in order to provide better understanding of yarn deformation in front of the projectile contact face. Four different viewing angles have been utilized, which show the presence of two potential deformation mechanisms that may responsible for the increased stress state in the yarn. First and foremost, as described in Chapters 4-7, there exists a local stress concentration at the corners of the projectile-yarn interface. Additionally, there is the presence of a yarn bifurcation process that seemingly tends to separate the yarn at $\sim45^\circ$ angles from the projectile contact face, thereby increasing the strain felt by the filaments as they move away from the projectile-contact site. Both of these mechanisms appear to be clearly evident with flat-faced projectile impact conditions, while only the latter is present for the round impact conditions. Thus this bifurcation behavior could be one the culprits promoting local failure of the yarn at velocities below the critical impact condition, even for seemingly stress-concentration free (at least from notches) round projectiles. Several impact conditions are discussed in detail and reference is made to further impact events that are shown in Chapter 6.
8.2 Introduction

In efforts to gain initial understanding of soft armor behavior when impacted at velocities above their respective ballistic limit ($V_{50}$), it becomes important to analyze the means of failure at the impact site, and the mechanism(s) in which energy is dispersed throughout the system. At velocities well below the $V_{50}$, a number of different energy dissipation mechanism play a role in halting the incoming projectile, be it friction, trellising, yarn tension, etc. [7] [8]. Interestingly, at velocities surrounding the $V_{50}$ of the system, local failure of the principal yarns becomes of greatest importance as is clearly demonstrated by the Cunniff parameter [49] and at even higher impacting velocities, various authors have stated that fabric fails in an inelastic fashion [7, 9, 126]. In the opinion of the current author, this inelastic failure causes the well known drastic loss in fabric energy dissipation capability at velocities just above the critical $V_{50}$ velocity due in part to the immediate yarn rupture which does not allow for any substantial elastic loading of the principal yarns. Surrounding fabric is thus deemed useless in moving energy away from the impact site. Evidence of such localized failure is also demonstrated by the lack of effect in fabric clamping aperture size, which is shown in Figure 8.10. At velocities above the $V_{50}$, projectile residual velocities from impacted fabrics with the differing boundary sizes converge, which the current author believe to be an effect of immediate inelastic failure, essentially rendering the clamping boundary of no consequence in the fabric performance, as there is insignificant time for wave propagation to ever meet the clamped region. Such an immediate failure is also seen in transverse impact of single yarns, and although not as common as full fabric experiments, is shown routinely throughout literature [50–61].

In such light, clearly there exists a strong importance to understand the effect of this immediate rupture and the amount of energy that is dissipated through this inelastic/immediate failure. Furthermore, it is also of interest to understand the eventual rise in energy dissipative capabilities of fabrics when impacted at velocities well above the $V_{50}$. Thus, previous impact data from Kevlar 29 fabric (8, 16, 18,
Figure 8.1. Effect of aperture size on the projectile residual velocity when impacting Spectra 215 fabric [7].
and 22 plies thick) using RCC projectiles (2-, 4-, 16-, and 64-grain) gathered from a report generated by Cunniff [12] has been used to develop a possible explanation for the energy dissipative capabilities of said fabric when impacted above the critical \( V_{50} \) velocity.

The traditional means of plotting the absorbed energy of a fabric system when impacted both above and below the \( V_{50} \) or ballistic limit is shown in Figure 8.2 and represents the impact of 0.28 kg/m\(^2\) Kevlar 29 fabric plies by a 16-grain RCC projectile. All listed data has been referenced from Cunniff [12]. Below the first peak in the energy absorbed versus strike velocity curve, the fabric effectively halts the incoming projectile, thus absorbing the entirety of the incoming projectile striker energy. At these sub-\( V_{50} \) velocities, the aforementioned energy dissipation terms (e.g. trellising, yarn pull-out, yarn tension, etc.) are activated. Interestingly, beyond the max energy peak, the energy absorbed by the fabric system experiences a sharp drop, indicating a sharp change in deformation and failure mechanism. Subsequently, the energy dissipation drop tends to a minimum, at which point the resulting absorbed energy begins to once again increase in a parabolic fashion with respect to the strike velocity. This monotonic increase in energy absorption with respect to impact strike velocity brings one to believe that the additional energy absorption mechanism must be related to the square of the projectile exit velocity. Such a relationship is further corroborated by normalizing the absorbed fabric energy by the striking energy, which is shown to approach a constant value at high impact velocities, demonstrated in Figure 8.3a. It is important to note that the data presented in Figure 8.3 represents shot data from Kevlar 29 fabric systems composed of either 8, 16, 18, or 22 ply layers. Each of these fabric systems was shot with a series of impact velocities ranging from \(~400\) m/s to \(~1400\) m/s using 2-, 4-, 16-, and 64-grain RCC projectiles. Again, as previously mentioned, each \( E_{abs}/E_{strike} \) curve approaches a constant value at high impacting velocities, albeit there is no immediate correlation between data using different shootpack thicknesses or differing projectiles.
Figure 8.2. Impact into Kevlar 29 fabric using a 16-grain RCC projectile. The fabric is 0.28 kg/m² and all listed data is from Cunniff [12].
Figure 8.3. Data from Figure 2 (16-grain), along with additional shot data using 2-, 4-, and 64-grain projectiles is normalized with respect to the impacting energy. (b) Normalizing all data by the projectile OD collapses different projectile sizes onto one another. (c) Normalizing by the system areal density collapses the different ply levels onto one another when impacted by similar projectiles. (d) Normalizing data using both the projectile diameter and the number of plies collapses all data cases onto one another.
Interestingly, by normalizing the data shown in Figure 8.3a by the projectile diameter \( D_{\text{proj}} \), data from similar thickness shootpacks collapse onto a similar curve, yielding the understanding that the absorbed energy must be some function of the projectile diameter. Furthermore, by normalizing the data shown in Figure 8.3a by the number of fabric plies, \( V_{50} \) data from different fabric thicknesses also collapses onto one curve when impacted by the same projectile type, which is shown in Figure 8.3c. This leads to the additional understanding that at high impacting velocities, the energy absorbed by the fabric can be partly described by some sort movement of the fabric within the system with the velocity of the impacting projectile. Finally, if both projectile diameter and the number of plies is used to normalize the data shown in Figure 8.3a simultaneously, all presented curves are collapse onto one master curve, yielding the assumption that projectile diameter and fabric areal density are two key parameters when comparing impact data from fabric systems composed of identical material (but are impacted with differing projectiles and contain different arial densities). It is now important to remember that both of these normalizations, \( D_{\text{proj}} \) and fabric thickness, are functioning on the absorbed fabric energy and are thus both directly related to the difference in squared velocity terms \( V_{\text{impact}}^2 \) and \( V_{\text{exit}}^2 \).

The question then arises as to why both projectile diameter and fabric thickness play such a powerful role in governing the resulting energy absorption capability of a fabric system when impacting at velocities above the \( V_{50} \) of the system. In this light, similar to work by Nguyen et al. [159] it is proposed that two main energy dissipation mechanisms govern system response at high projectile impacting velocities, namely (1) local yarn failure, and (2) acceleration of fabric material up to the exit velocity of the projectile. Additionally, it is proposed that this two-phase process can happen in either of two modes (1) shear plugging or (2) local ‘tensile’ rupture (‘tensile’ is used because the yarns could fail in one location in front of the projectile head, not forming a shear plug, but the actual means of failure in the filaments themselves is not limited to pure tension). These two modes more summarized by the following:

1. Shear plugging (demonstrated in Figure 8.4(a))
Figure 8.4. Schematic of local failure process either as (a) shear plug formation and subsequent acceleration of the plug up to the exit velocity of the impacting projectile, or (b) acceleration of a local zone around the impacting projectile followed by local failure, leading to full perforation.

(a) Local yarn failure mode: shear plug formation

(b) Fabric acceleration: accelerate shear plug up to exit velocity of projectile

(before and after shear failure)

2. Local ‘tensile’ rupture (demonstrated in Figure 8.4(b))

(a) Local yarn failure mode: perforation of fabric via singular location yarn failure

(b) Fabric acceleration: accelerate fabric within affected zone up to the exit velocity of the projectile before fabric perforation

The presence of a local shear formation is described in literature [7,9,126], and fits within the inelastic rupture mode of the fabric system that was previously described. The energy required to generate such a shear plug is described by the following:
\[ E_{\text{shear}} = (\beta \pi D_{\text{proj}^t})(\frac{1}{2}\tau_{\text{max}}\gamma_{\text{max}}) = \text{constant} \] (8.1)

where \( D_{\text{proj}^t} \) and \( t \) represent projectile diameter and fabric thickness, respectively. \( \beta \) is a spatial parameter which adjusts the size of the shear plug, being roughly 1.3-1.5 [54,135]. Thus the entire first term of Equation 8.1 represents the circumferential area forming the outer diameter wall of the shear plug. \( \tau_{\text{max}} \) and \( \gamma_{\text{max}} \) represent the composite/fabric through thickness shear strength and failure shear strain, respectively. In the present work this term is assumed to be rate insensitive, with the amount of energy forming for a specified fabric system being solely a function of the size of the shear plug when changing projectile size or fabric thickness. Following shear plug formation, it is then assumed that the shear plug must be accelerated up to the exit velocity of the impacting projectile, requiring an energy value described by the following:

\[ E_{\text{plug}} = \frac{1}{8}(\rho_{\text{fabric}})(\beta \pi D_{\text{proj}^t}^2)V_r^2 \] (8.2)

where \( \rho_{\text{fabric}} \) and \( V_r \) represent fabric density and projectile residual velocity, respectively. Thus, it is assumed that during high-velocity impact of fabric systems, specifically above the fabric’s ballistic limite, Equations 8.1 and 8.2 can describe the majority of the resulting energy dissipation. Figure 8.5(a) demonstrates both energy terms separately. The plug acceleration term is seen to very closely follow the shape of the actual experimental energy absorption curve, but with what appears to be a constant difference at high strike velocities. Thus, it is believed that this constant difference portion can be described by the inelastic plug formation, being a term that is roughly estimated from previous literature and then is fitted so fall on the actual experimental curve presented in Figure 8.5(b).

Most importantly, once this shear plug formation term is fitted to one specific data subset, \( \tau_{\text{max}}\gamma_{\text{max}} \) is left unchanged for the remaining 15 experimental conditions with only the size of the projectile and the fabric pack thickness to be changed. Remark-
Figure 8.5. (a) The two energy terms described by Equations 8.1 (shear plug formation) and 8.2 (shear plug acceleration) are described by red asterisk and red circles, respectively. (b) Both energy terms are summed and plotted with the actual experimental energy dissipation measured from experiment. All data is from 18-ply packs shot with 4-grain projectiles.
ably, using these two simple energy terms to predict the energy absorption capability of the fabric, an extremely good fit is found. In Figure 8.7 is plotted the difference between predicted energy absorption and the actual fabric energy absorption. At high velocities, for all 16 test cases, the resulting difference approaches zero, indicating that the two energy terms account extremely well for the energy dissipation mechanism of the fabric at high impacting velocities. Most surprisingly, the energy term containing the acceleration of the shear plug becomes of most importance at high impacting velocities, which makes reasonable sense as the curves shown in Figure 8.3a demonstrate that at high impacting velocities, the energy dissipated by the fabric normalized by the impacting energy becomes a constant leading one to believe that the energy absorbed by the system must somehow be a function of the square of a velocity term.

Although not displayed in the current findings due to similarity, analysis of the second means of local failure is follows along the same lines as the shear plug failure mode, with two main energy terms governing the energy dissipation from the projectile during impact. Firstly, the fabric direction surrounding the footprint of the
projectile must be accelerated up to the exit velocity of the projectile. Secondly, local perforation of the principal yarns occurs in front of the projectile. It must be noted that both energy terms are very similar in nature to the shear plugging mode, except for the order of their physical occurrence. In the local tensile failure mode, fabric is first accelerated to the projectile velocity, followed by local fabric perforation. In the shear plug failure mode, initial failure of the fabric occurs along with simultaneous acceleration of the local fabric of the in and around the shear plug up to the velocity of the projectile. An example of the local tensile failure mode is demonstrated in Figure 8.7. In this failure sequence, the fabric clearly undergoes the local tensile failure deformation.

Due to the extremely detrimental impact of failure occurring both locally and rapidly around the impact sight, thereby drastically reducing the energy absorbed by the fabric system, it is of great interest to understand the actual means of failure about the head of the projectile, specifically, how and where the constituent yarns indeed fail. In order to reduce the complexity of the impact event to as basic a level of possible, single yarn transverse impact has been performed using two different projectile nose geometries and two types of yarn. Effort has been placed on visualizing the impact event and as such, large projectile diameters have been used to so to increase the viewing area during the impact event.

8.3 Experimental

Similar to previous impact experiments shown in Chapters 6 and 7, the body of work utilized a powder-breatch smooth bore gun to fire 20 mm projectiles. A basic schematic of the powder gun system can be seen in Figure 8.8. Upon exiting the bore of the gun system, the projectile passed through a three laser diode system, which was used to determine the projectile impact velocity. Three different steel projectile geometries were utilized, each with an 18 mm diameter so as to fit within a 20 mm polyurethane sabot. An image of all three projectiles can be seen in Figure
Figure 8.7. Kevlar IIIA ballistic panel impacted with a 170 grain 30-06 projectile at 733 m/s. Frame spacing is 0.5 μs. Fabric demarcations represent 1 cm spacing.
Figure 8.8. Basic schematic of gun and dual camera system used for imaging single yarn transverse impact experiments.

8.9 depicting a flat sharpened disc, a round head (half cylinder turned on the side), and a flat disc with a turned chamfer. This variety in projectile nose geometry was selected so as to gain additional understanding of the mode of localized failure during yarn-projectile contact.

Three different types of yarn were used for experiment, namely Kevlar® KM2, Dyneema® SK76, and AuTx®. The first two material types have been selected due to their heavy use in currently employed body armor systems, while the latter has been selected due to its possible future use in body armor. An example of typical physical and mechanical properties from these three material types can be seen in Table 6.2.

The main purpose of this experimental data set is visualize the impact of yarn using high-speed imaging so as to elucidate the mode of failure, with specific interest placed on determining the actual failure location. As such, several different yarn-camera orientations have been utilized, which are shown in Figure 8.10. Yarns have been hung in both vertical and horizontal geometries, and camera positioning has been both perpendicular and 47.5° from the projectile shot line. Both sets of yarn orientations and camera orientations have yielded four different vantage points to view
Figure 8.9. Large projectiles used for single yarn transverse impact experiments.
the contact of the projectile with the target yarn. Figure 8.10(a) demonstrates the
typical yarn camera geometry used throughout literature, wherein the camera view-
ing angle is perpendicular to both the shot line and the yarn longitudinal direction.
This viewpoint provides easy analysis of constituent filament failure, as failed mate-
rial will not travel with the transverse wave speed predicted from the Smith theory.
Figure 8.10(b) demonstrates a slight change in the camera viewing angle, in that it
has been rotated to $47.5^\circ$ from the projectile shot line. This allows for viewing of
local deformations occurring both in the plane perpendicular to the projectile shot
line and out of plane deformation in the direction of the projectile. Figure 8.10(c)
demonstrates a very stark change in the typical means of viewing yarn deformation in
that the yarn has been rotated $90^\circ$ about the projectile shot line so as to be concentric
with the camera viewing direction. This vantage point allows for visualization of the
local yarn cross-section during the impact event and also allows for detection of any
‘bouncing’ that may occur during the initial projectile-yarn contact. Figure 8.10(d)
demonstrates a slight change from the setup of Figure 8.10(c) in that the camera has
been rotated to be $47.5^\circ$ from the projectile shot line. This viewing angle allows for
easy access into yarn cross-section mechanics such as potential spreading of the yarn
or even bifurcation during contact.

For yarn orientations shown in Figures 8.10(a) and 8.10(b), system alignment was
performed as described in Hudspeth et al. [160]. Specifically, a plumb bob was used
to align the yarn clamps in a vertical fashion and bubble-level was used to provide
horizontal alignment to the gun barrel. Locating the yarn with respect to the center
of the projectile shot line was performed using a laser bore-site. Yarn orientations
described in Figures 8.10(c) and 8.10(d) were aligned using measurements from frame
standoffs were squared with framing placed in the bed of the target chamber. As with
the vertically oriented yarn, the horizontal yarn was located within the projectile
shot line via laser bore sight. Finally, camera alignment was performed via arial
photography, taken in order to include barrel, yarn, and camera assemblies.
Figure 8.10. Camera orientations used to image transverse impact into single yarns. (a) Vertically suspended yarn with camera perpendicular to shot line, (b) Vertically suspended yarn with camera 47.5° from shot line, (c) Horizontally suspended yarn with camera perpendicular to shot line, (d) Horizontally suspended yarn with camera 47.5° from shot line.
Camera triggering was performed in a similar fashion to that described in Hudspeth et al. [160], wherein the rising edge of the detected longitudinal stress wave at the yarn load-cell/end-clamp was used to promote image storage. As it takes several microseconds for the longitudinal stress wave to travel from the impact sight to the end clamp, a post-trigger feature was used in order to capture images before the actual stress-wave detection, thereby ensuring captured images preceded the projectile-yarn impact event. For the vertically oriented yarn (Figures 8.10(a) and 8.10(b)), yarn ends were held between clamps with the rest of the material residing in open free space. In contrast, for the horizontal yarn (8.10(c) and 8.10(d)), yarn ends were wrapped $90^\circ$ on an arc path line about 50 mm diameter steel posts and then into the load cell clamps. Such a wrap technique was used in order to remove the clamping regions from the field of view of the camera and illumination source, with careful attention paid to the orientation shown in Figure 8.10(c), as any fixture in line with the yarn length would inherently block the desired image path.

8.4 Results and Discussion

Albeit quite different from the rest of this document, this specific chapter is more directed at providing visual interpretation of the failure event occurring in front of both the round projectile head and the flat projectile head. Specific interest has been placed at determining the location of failure at velocities above the experimental critical velocity transition regime, yet below the theoretical critical velocity. It is inside this region where failure should theoretically not occur, yet does, and thus it is of interest to determine how and where failure does indeed progress. Reasoning for understanding the location of failure stems highly from the idea that ‘fiber bounce’ promotes the premature failure of the yarn. Understanding this location of failure inevitably leads to better understanding of potential failure modes. Description of the various shot conditions can be found in Table 8.1 as most shot conditions will not be described due to brevity to the current document.
Table 8.1. Experimental conditions for transverse impact of 20 mm projectiles into Kevlar® KM2 and Dyneema® SK76

<table>
<thead>
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<th>Shot ID</th>
<th>Material</th>
<th>Projectile Geometry</th>
<th>Shot Velocity (m/s)</th>
<th>Camera Orientation</th>
<th>Figure ID</th>
</tr>
</thead>
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<td>flat</td>
<td>552</td>
<td>(a) vert - 0</td>
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<td>963</td>
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<td>SK76</td>
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<td>891</td>
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</tr>
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</tr>
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<td>646</td>
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</tr>
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<td>887</td>
<td>(c) horiz - 0</td>
<td>Figure A.6</td>
</tr>
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<td>SK76</td>
<td>flat beveled</td>
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8.4.1 Kevlar KM2 - Flat Projectile

Inside Critical Velocity Transition Regime

The most typical means of viewing the deformation process exhibited by high-performance yarn subjected to transverse impact can be seen in Figure 8.11, wherein the projectile flight path is 90° perpendicular to both the camera viewing angle and the suspended yarn. This orientation can be found in Figure 8.10(a), seemingly demonstrating an appropriate angle to verify Equations 6.1-6.5. Such example of this process has been used through Chapters 4-7. An example of this perpendicular viewing angle can be seen in Figure 8.11 wherein a KM2 yarn has been impacted at a velocity of 624 m/s. Although difficult to see in this specific image sequence, which is facing an impact condition very close to the upper end of the critical velocity transition region, it is reasonably clear to see that the failure process is slightly progressive in nature with initial failure developing at the corners of the projectile. Interestingly, later times show almost a feathering of the yarn directly in front of the projectile. The reader is directed at image sequences in Chapter A for impact conditions at both lower and higher velocities (as well as with SK76 yarn instead of KM2). Note that higher impacting velocities more clearly show failure at the corners of the projectile.

In order to gain additional understanding of the impact process, an additional camera was used to image the impact event, with the new camera being 47.5° from the projectile shot line, thereby giving a face view of the projectile-yarn contact behavior. This viewing angle is described in Figure 8.10(b) and an example of an impact image sequence can be seen in Figure 8.12, wherein KM2 yarn has been impacted with a flat projectile at a velocity of 624 m/s. This is the exact same impact event that is shown in Figure 8.11. In this viewing angle, it much easier to see the yarn behavior in front of the projectile, which interestingly seems to grow in cross sectional area. It is suggested that such an expansion is allowed due to partial yarn failure, and the resulting contact conditions that develop between the cylindrical yarn and the flat projectile face. Although better demonstrated in Figure 8.13 and even better in figures.
Figure 8.11. 07281516a. Frame separation is 200 ns.
contained within Chapter A for higher shot velocities, the yarn actually bifurcates and two lobes are sent moving in ∼45° angles away from the projectile face. It is suggested that along with the local stress concentration of the projectile corner, this bifurcation may add additional straining to the yarn, especially to filaments near the outskirts of the bifurcating lobes. Again, in the current image sequence of Figure 8.12, the yarn looks to be feathered to the nature of the partial failure.

In order to gain further understanding of the yarn cross-section during impact, especially the aforementioned bifurcation process, the yarn was rotated 90° so as to be suspended horizontally, which is demonstrated in the schematic provided in Figure 8.10(c). An example of this image sequence can be found in Figure 8.13, wherein a KM2 yarn has been impacted at velocity of 624 m/s, which again is within the upper region of the transverse critical velocity described in Chapter 6. As such, the failure of the yarn is not as clean as demonstrated in alternate impact sequences that are depicted in Chapter A. Yet the yarn cross-section can be clearly seen in Figure 8.13. There does appear to be the yarn bifurcation, even within the transverse critical velocity transition regime, and is demonstrated in two modes. First, and probably the most revealing, is the shape of the shock front that is emanated ahead of the yarn upon acceleration, which most probably occurs due to yarn bouncing (the compressive stress wave moves through the cross-section of the yarn and reflects back as a tensile wave). It should be noted that the bounce behavior should be incremental, occurring for each and every filament in a sequential form. As the compressive stress wave moves through one filament to the next, filaments that don’t have a move forward neighbor will see their velocity increase (potentially double) with respect to the projectile, and forward away from their previous neighbors. This then allows the previous neighbors to experience no forward neighbor, thereby allowing the new forward most contacted filaments to accelerate ahead of the projectile. This then cascades down to the remaining filaments ahead of the projectile. This increased velocity of the filaments with respect to the projectile thereby causes the shock which can be seen in Figure 8.13. Notice again the shape of the shock, it appears to have a
Figure 8.12. 07281516b. Frame separation is 200 ns.
two-lobed structure, revealing that the yarn filaments must be moving as two general bodies in different directions (∼45° from the yarn face). The bifurcation of the yarn can also be seen by the yarn itself. Notice in Figure 8.13 how later stages of the impact even show a yarn cross-section that seems to have somewhat separated into two slightly different lobes. Note that these lobes in the later times most likely represent failed yarn, while any intact yarn would still be closer to the impact face. Again it is noted that more clear views of this bifurcation phenomenon can be seen in image sequences shown in Chapter A.

Finally, as with the first impact experiment shown in Figures 8.11 and 8.12, it was deemed of use to implement a secondary camera located at an angle to the projectile shot line (47.5°), in order to gain additional understanding of the deformation of the yarn in front of the projectile impact face. A general schematic of this experimental geometry can be seen in Figure 8.10(d) and a representative image sequence of an impact event can be seen in Figure 8.14. Figure 8.14 originated from the same impact event as to that shown in Figure 8.13, namely, a KM2 yarn impacted with a flat faced projectile moving with a velocity of 622 m/s, perpendicular to the yarn longitudinal direction. As with Figure 8.12, the geometry of the yarn again looks to exhibit a large degree of expansion directly in front of the projectile face. This expansion, along with the local stress concentration at the projectile corners, is believed to be responsible for the increased stress-state that is felt in the yarn. Again, the yarn looks to have been feathered due to the impact event. It is also interesting to note that a transverse wave can be clearly seen to move inwards from both corners of the projectile at early times post impact.

Above Critical Velocity Transition Regime

It is now of interest to briefly discuss the impact of Kevlar® KM2 above the experimental critical velocity, yet below the theoretical critical velocity. As with the previous discussion provided for KM2 impact with a flat faced projectile within the
Figure 8.13. 07291503c. Frame separation is 200 ns.
Figure 8.14. 07291503d. Frame separation is 200 ns.
critical velocity transition region, specific attention has been placed on determining both the location and mode of failure directly in front of the projectile head. As such, the four camera geometries listed in Figure 8.10 have been utilized.

An image sequence of the failure process of a KM2 yarn impacted at a velocity of 851 m/s using the typical camera geometry, which is shown in Figure 8.10, can be seen in Figure A.18. Note the clear demonstration of failure at the at the edges of the contact zone between the yarn and the projectile face. It appears that material in front of the yarn is carried along with the projectile post failure at the projectile corners. Additionally, note the lack of any appreciable transverse wave development on either edge of the contact zone. From this vantage point and at this impacting velocity, a high level of confidence is felt regarding the belief that failure is occurring at the projectile corners.

In order to gain additional insight on the movement of the failed yarn in front of the projectile head, the alternative viewing angle shown in Figure 8.10(b) has been implemented. An image sequence from this viewing angle can be seen in Figure 8.16, wherein a Kevlar® KM2 yarn was shot with the flat faced projectile at an impacting velocity of 837 m/s. It is important to note that the typical viewing angle (Figure 8.10(b)) for this exact shot can be seen in Figure A.23. In Figure 8.16 failure is once again apparent on the corners of the projectile, but the deformation process appears to follow a splaying sequence, similar to that previously described. Due to early failure of the filaments at the corner of the projectile, the splayed yarn appears to take a bifurcated path with roughly half of the yarn moving perpendicular to both the shot line direction and the yarn longitudinal direction (i.e. along the face of the projectile). Finally, it is important to note that the sharp failure at the corners of the projectile is not as stark as that shown in Figure A.18, as the impact event described in Figure 8.16 utilized a projectile with bevelled corners, as compared to the sharped edge flat projectile that was used for the impact event recorded in Figure A.23.

In efforts to better understand the splaying process, the experimental geometry described in Figure 8.10(c) has been used to look at the yarn cross-section in front of
Figure 8.15. 06231527a. Frame separation is 200 ns.
Figure 8.16. 07281515b. Frame separation is 200 ns.
the flat-faced projectile head. An image sequence using this experimental geometry can be seen in Figure 8.17 for a Kevlar® KM2 yarn impacted at 845 m/s. Upon contact of the projectile with the yarn, a double-lobed shock front is developed, demonstrating the increased velocity of the yarn with respect to the projectile and the bifurcation or separation of the yarn shown in Figure 8.16. Additionally, it can be seen from the image sequence described in Figure 8.17 that the yarn cross-section undergoes a flattening/splaying seemingly moving the two halves of the yarn at a slight angle from the projectile face (one moving upwards and the other moving downwards in the reference frame shown in Figure 8.17).

Finally, in order to gain additional understanding of the yarn cross section during the failure process, the camera viewing angle was transitioned to the viewpoint shown in Figure 8.10(d) and an image sequence from Kevlar® KM2 impacted by a flat projectile at 822 m/s is shown in Figure 8.18. Note the large expansion of the yarn that progresses after impact. It must be noted that such a drastic ‘expansion’ is an artifact of the failure process; due to tension loss from filament failure, broken filaments are able to expand the yarn cross-section as they are not constrained from yarn material outside the contact zone.

8.4.2 Transverse Impact ‘Bouncing’

Finally, it is of interest to point out the presence of bouncing in front of the projectile head during transverse impact. Figure 8.19 depicts an AuTx yarn being impacted by a 30-cal FSP at 480 m/s. Although somewhat difficult to see due to spatial resolution limitations of the camera system, bouncing of the AuTx yarn in front of the flat portion of the projectile head is clearly evident. Initial yarn-projectile impact appears to occur between at the 0.2 μs, at which point the yarn directly ahead of the projectile increases in velocity. Simultaneously, transverse wave forms emanate from both corners of the projectile, moving both towards and away from the middle of the projectile centerline. The two inward moving waves propagate towards the
Figure 8.17. 07311504c. Frame separation is 200 ns.
Figure 8.18. 07291502d. Frame separation is 200 ns.
center of the projectile from initial contact (∼0.2 μs) until they reach one another at roughly 2.2 μs (thus a transit time of ∼2 μs), agreeing very well with the theoretical transverse wave speed of 938 m/s, which predicts a transit time the FSP corner to the middle of the projectile as 1.9 μs. It is important to note that the material behind the transverse stress waves is slowed. Further travel of the inward moving stress waves subsequently slows the forward velocity of the yarn, ultimately allowing the projectile to ”catch up” and make subsequent contact. In this image sequence, the yarn does not appear to undergo a second contact condition promoting an additional bounce.

8.5 Conclusions

Single yarns have been impacted with 20 mm round and 20 mm flat projectiles, with an emphasis placed on determining the location and mode of failure in the yarn. It has been determined that above the critical velocity, yarn failure in the flat faced projectile occurs at the corners of the projectile head. Within the critical velocity, inherently only partial yarn failure occurs, but failure is still believed to have occurred at the corner of the projectile. Reasoning for failure of the filaments below the transverse critical velocity is believed to occur for two reasons, namely the stress concentration occurring at the corners of the projectile, and also due to yarn expansion that occurs from bifurcation of the cylindrical cross-section. Impact with the round projectile heads has also caused failure quite similar in nature to the flat projectile (bifurcation), but the failure location seems to be random throughout the entire yarn-projectile contact zone. No specific region of failure can be identified, except to be stated that failure occurs somewhere (seemingly random) between the yarn-projectile contact face. Regardless, both impact conditions do indeed require further analysis, preferably via full scale yarn simulation, which is much above the expertise of the current author. Although local stress states are demonstrated in Chapter 6 for a single filament impacted with a projectile, full scale yarn impact requires better
Figure 8.19. Transverse impact of an AuTx yarn with a FSP at 480 m/s.
understanding of contact conditions present during the impact event and a powerful enough computation system to analyze 500-1000 single filaments simultaneously.
MULTI-AXIAL FAILURE OF HIGH-PERFORMANCE FIBER DURING
TRANSVERSE IMPACT

VOLUME 2

A Dissertation
Submitted to the Faculty
of
Purdue University
by
Matthew C. Hudspeth

In Partial Fulfillment of the
Requirements for the Degree
of
Doctor of Philosophy

May 2016
Purdue University
West Lafayette, Indiana
9. High speed synchrotron x-ray phase contrast imaging of dynamic material response to split Hopkinson bar loading

Adapted from:

9.1 Abstract

The successful process of amalgamating both the image capabilities present at the APS beamline 32ID-B and the proficiency of high-rate loading offered by the Kolsky compression/tension bar apparatus are discussed and verification of system effectiveness is expressed via impact on various material systems. Single particle sand interaction along with glass cracking due to impact have been analyzed via the compression setup, while fiber-epoxy interfacial friction, ligament-bone debonding, and single-crystal silicon failure was achieved within the described tension scheme. Analysis of said material systems being loaded with the Kolsky bar apparatus while visually tracking deformation history via Phase Constrast Imaging demonstratively depicts the effectiveness of the novel union between these two powerful techniques, thereby allowing for in-situ image analysis *inside* of the material system during loading rates commonly encountered in blast environments.

9.2 Introduction

Characterization of materials at elevated strain rates is of dire importance when analyzing structural response in any sort of impact or ballistic application. It is well
known that numerous materials exhibit a loading rate sensitivity, thereby rendering typical quasi-static analysis inadequate. Thus, the use of high-rate testing methods has become crucial in material response study, with the implementation of the Split-Hopkinson bar [161] and gas gun loading [162] being the most common and robust techniques when testing medium ($10^2 - 10^5 \text{ s}^{-1}$) and high ($10^5 - 10^9 \text{ s}^{-1}$) strain rate loadings, respectively. Historically, Kolsky bar research has been limited to determining sample behavior via voltage signals generated from strain gages attached to both the incident and transmission bars or from embedded quartz crystals, while gas gun loading measurements have been dominated by surface or interface measurements with optical velocity or displacement interferometry [163, 164].

With the advancement in high speed imagery, many works have began analyzing the deformation history of the material response via some variant of a surface structure tracking scheme. This of course requires the assumption that response within the sample is similar to the surface particle deformation, and unless the material itself is transparent, analysis of sample through-thickness deformation is impossible. Furthermore, much development is currently being directed at designing materials which fail in an extreme energy dissipative manner, requiring researchers to understand the actual failure phenomenon presented by a material system. For example, in a penetration environment, while the armor material’s stress-strain response is of great importance, much attention is being placed on the defeat mechanisms presented by the material and the ultimate energy dissipation history developed by the armor structure as a whole. If an armor system can disperse incoming penetration energy more quickly and effectively, new threat halting capabilities may be realized. In light of this, it is of upmost importance that specific material failure phenomenon be better tracked and understood in loading environments representative of actual use.

Recently, “white beam,” single pulse, Phase Contrast Imaging (PCI) has been realized at the APS 32ID-B beamline under gas gun loading allowing for shock-wave propagation analysis within the material body [165]. It is the goal of this study to adapt this erudite technique to the commonly employed Kolsky bar apparatus.
in efforts to track material behavior when loaded to strain rates typical in blast environments.

9.3 The Kolsky bar PCI experiment setup at the APS beamline 32ID-B

The complex index of refraction of a material for x-rays is \( n = (1 - \delta) + i\beta \), where \( \beta \) and \( \delta \) account for absorption, and phase shift, respectively. Let the sample–detector or scintillator distance be \( z \). For contact image radiography or absorption imaging, \( z = 0 \), while for propagation-based phase contrast imaging \([166–170]\), \( z \) is finite and in the near-field Fresnel region (the Fresnel number \( \gtrsim 1 \)). A heterogeneous phase object (small \( \beta \)) induces spatial variations in the phase of x-rays, \( \psi(x, y, z = 0) \), and local curvature in the transmitted wavefront. During free space propagation from \( z = 0 \) to \( z_0 \), overlap and interference of the wavefront modulate the intensity. \([169, 170]\]

The intensity change due to this propagation is proportional to the Laplacian of \( \psi(x, y, z = 0) \), yielding edge enhancement. \([169]\) This is the benefit of PCI for resolving inhomogeneities and structure changes in low-\( Z \) (atomic number) materials or phase objects which are difficult to measure with contact image radiography. X-ray PCI also has certain advantages over conventional optical imaging in penetration depth and spatial resolution (the high scattering of visible light is a main drawback).

In principle, PCI requires spatially and spectrally coherent x-rays, but the requirement of spectral coherence may be relaxed considerably. \([168]\) High coherence, high flux, and high repetition rate synchrotron undulator sources with multiple harmonics (“white beam”) are advantageous for dynamic PCI without a flux loss due to monochromators.

Dynamic white beam PCI measurements under the Kolsky compression/tension bar loading has been performed at the APS beamline 32ID-B. The x-ray beam at 32ID-B employs APS Undulator A with a period of 3.3 cm and length of 2.4 m. The specimen was located approximately 40 m away from the undulator light source. The actual beam spot size (typically 1–2 mm in each direction) on the sample was
controlled with adjustable slits in both directions. It was possible to vary the x-ray spectrum characteristics as desired via changing undulator gap. For example, a gap change from 30 mm to 20 mm in the standard APS operation mode can lead to shifting of the first harmonic from 13 keV to 9 keV, accompanied by an increase in photon number by a factor of 3. [165]

Dynamic loading and event duration determine the temporal resolution and number of frames required for dynamic PCI measurement and set the requirements for time structure and flux of synchrotron pulses and for detectors (single frame exposure time, framing rate, and response of scintillators). For Kolsky bar experiments, the
event duration is on the order of 100 μs; while single-pulse temporal resolution [165] is desirable, a temporal resolution of 100 ns−1 μs can be adequate. Here, the temporal resolution refers to the x-ray integration of time for a single frame, not the frame separation (framing rate).

The time structure of x-ray pulses or the corresponding electron bunches depend on the operation modes of the APS electron storage ring. The pulse train is circular, and one revolution of electrons takes 3.68 μs. The number of photons in an x-ray pulse scales linearly with bunch current. For time-resolved measurements, it is imperative to consider the bunch/superbunch current (photon number), width (temporal resolution), and separation (framing rate), as well as synchronization of an x-ray bunch/superbunch with a dynamic event and time constants of scintillators and detectors. For Kolsky bar experiments, the hybrid fill mode appeared to be the most appropriate (Figure 1). In the hybrid mode, a single bunch containing 16 mA (singlet) is isolated from the remaining bunches by symmetrical 1.594 μs gaps (\(A\hat{B}\) and \(A\hat{C}\), Figure 1). The remaining current is distributed in 8 groups of 7 consecutive bunches (septuplets) with a maximum of 11 mA per group, a periodicity of 68 ns, and a gap of 51 ns between groups (\(G\hat{H}\)). [171] The total length of this bunch train is 500 ns (\(B\hat{C}\)). The hybrid mode is denoted as “1 + 8×7,” where “1” stands for the singlet, and “8×7,” refers to 8 groups of septuplets which are collectively referred to as a superbunch. If the singlet in the hybrid mode is used for PCI, the temporal resolution is the pulse width (<100 ps), pulse separation (framing rate) is 3.68 μs, and bunch current is 16 mA, while for the superbunch PCI, their counterparts are 500 ns, 3.68 μs and 88 mA. These numbers are <100 ps, 153 ns, and 4.25 mA for single-pulse PCI in the standard or 24-bunch mode. The superbunch of the hybrid mode has been utilized in these Kolsky bar experiments. Compared to the singlet PCI in the hybrid and standard modes, temporal resolution, thus framing rate, has been reduced, but significant gains have been achieved in photon number and detection.

The dynamic loading device employed was the split-Hopkinson or Kolsky bar and the design, testing, and applications of such bars have been detailed by Chen and
Song. [161] For the APS PCI experiments, two miniature Kolsky bars were specifically developed, allowing for both tension and compression loading schemes (Figs. 2 and 3). Due to space constraints of the X-Ray containment hutch, both bar systems were miniature in nature, thereby removing the transmission bar for each setup. As the purpose of this study was to establish the efficacy of employing dynamic white beam PCI measurements under Kolsky compression/tension bar loading, this modified design was accepted with note that in order for valid stress-strain response analysis, the impedance mismatch between the bar-end and sample must be extremely large, thereby requiring samples to be of very fine cross-sectional area.

Each bar was outfitted with its own barrel, $\frac{1}{2}$-in striker, and $\frac{1}{2}$-in incident bar, but for ease of experimentation, they shared the same gas tank, firing mechanism, and baseplate assembly. The firing system consisted of a zero-pass solid-state relay, which was triggered by a 3V input DC current, thereby allowing a 120V AC current to flow into a fast-response solenoid valve. This valve was located close to the inlet nozzle of each barrel assembly, allowing for the minimal ramp-up pressure needed to ensure consistent delay timing from solenoid trigger to striker-incident bar impact. Two semi-conductor strain gages were attached to the surface of the incident bar of both setups and were wired in a two-arm active wheatstone bridge, thereby aiding in correction for slight bending present in the bar alignment. Timing differences between solenoid valve opening and strain gage activation were obtained as a function of air pressure, thereby allowing for the necessary accurate determination of the timing sequence in dynamic measurements as shown in Figure 6.

A fast response quartz load cell was mounted onto an isolated rigid fixture, ensuring minimal load cell deflection. Selection of load cell capacity was then carefully chosen for each testing scheme in order to provide appreciable excitation during the loading history, while simultaneously ensuring prevention of response overload. Both load cell and strain gage voltage signals were concurrently collected with an oscilloscope, and representative histories can be seen in Figure 5. These signals were further synchronized with the PCI images (Figure 5) in efforts to correlate visual material
deformation history with the recorded loading and bar velocity response. It is also important to note that the strain gage signal was further used to both trigger a PCI camera and to determine the bar-end velocity history undergone during the impact event. Using the voltage signals generated by the strain gages from both the tension and compression bars and 1D-wave mechanics, typical incident bar velocities were determined to be 4-6 ms$^{-1}$ and 4-8 ms$^{-1}$, respectively, for charging gage pressures measured between 10-20 psi, being determined by the material of interest. Furthermore, all described system assemblies are demarcated in Figs. 1-3 in efforts to aid in understanding of both the loading and imaging scheme relative geometries.

A single crystal scintillator Lu$_3$Al$_5$O$_{12}$:Ce or LuAG:Ce (Crytur Ltd., Czech Republic) was used to convert x-rays to visible light. The dimensions of the implemented scintillator were 10 mm $\times$ 10 mm $\times$ 100 $\mu$m; its decay time was about 45–55 ns, and the emission spectrum peaked at 530 nm. [165,172] Both the scintillator decay time and its exposure time to the 8 $\times$ 7 superbunch (500 ns) were considerably shorter than the interframe separation ($n \times 3.68$ $\mu$s, where $n$ is an integer varying from 1 to 5 in these experiments), so the ghost image effect [165] due to the preceding exposure was deemed negligible in current exposure. The high-speed camera for capturing op-
Figure 9.3. Photographs of the experimental setup at the APS beamline 32ID-B, viewed approximately along (a) the x-ray beam direction, and (b) the bar direction. 1: compression bar; 2: tension bar; 3: strain gage on the compression bar; 4: load cell or force transducer; 5: scintillator and optics; 6: high speed camera; 7: gas holding tank; 8: solenoid valve; 9: projectile-passing diode laser; 10: He-filled x-ray beam transporting tube (the arrow indicates the x-ray beam direction); 11: slow shutter; 12: gun barrel for the compression bar.

The critical images used was a Photron Fastcam SA1.1 with a 12-bit CMOS sensor. The full frame capabilities of this imaging system had 1024×1024 pixels, and the pixel size was 20 μm. Its global electronic shutter could operate at 1 μs, independent of frame rate. The frame rate could be increased from 5,400 fps to 675,000 fps with reduced image sizes, e.g., 1024×1024 pixels at 5,400 fps, 320×128 pixels at 100,000 fps, 128×64 pixels at 300,000 fps, and 64×16 pixels at 675,000 fps. The images were stored in memory for an event duration of >1 s. This camera could be phase locked or synchronized to an external source and triggered by an external TTL signal.

Schematic and photographs of the Kolsky bar PCI experiment setup are shown in Figs. 1 and 3, including the x-ray source, slits and control shutters, Kolsky bar loading system, and detector system. The undulator gap was typically 30 mm in these experiments, thereby allowing the source to deliver white beam x-rays with the fundamental centered at ~13 keV. The shutters were necessary due to the excessive
Figure 9.4. Air pressure vs. $\Delta t_{-3,0}$ calibration curves for the Kolsky (a) tension and (b) compression bars. $\Delta t_{-3,0} = t_0 - t_{-3}$, is the time interval between when the solenoid valve is triggered to open ($t_{-3}$) and when the strain gage is activated ($t_0$); also see Figure 6. The solid curves denote fitting with a power law.

Figure 9.5. Oscilloscope recordings of the strain gage and load cell (force transducer) signals. The camera is triggered indirectly with the strain gage signal. The strain gage excitation defines $t_0$ in Figure 6.

exposure to the high flux, white beam x-rays which were capable of damaging the sample and optics downstream. The Kolsky bar was oriented at a 90° angle with the x-ray beam, and a sample was placed in the x-ray beam path, between the bar end and a load cell (force transducer). The load cell would typically be replaced with a transmission bar, but there did not exist sufficient space in this hutch to
Figure 9.6. Timing/synchronization schemes for the Kolsky bar PCI experiments. DG-SS: single shot signal sent from a DG535 digital delay/pulse generator; sync: synchronize or synchronization; \( P_0 \): radio frequency or RF timing pulses (master clock), separated by 3.68 \( \mu s \) (271.7 kHz).

To accommodate such an assembly. Upon firing the system, x-rays were allowed to pass through multiple 2D slits and shutters, thereby impinging upon the sample which was concurrently undergoing dynamic loading. After exiting the sample assembly, the x-rays were finally passed through a downstream scintillator, being located 18 cm and 23 cm away from the axes of the tension bar and compression bar, respectively. Via the scintillator, the x-rays were then converted into optical photons, which were relayed to the high speed camera by a 45° mirror, a 10× microscope objective, and a tube lens which is detailed elsewhere. [165] The scintillator and optics were housed in a solid aluminum case and their position could be remotely adjusted with micron precision to achieve best focus, and the whole detector system was set on a remotely-controlled translation stage.

For a successful dynamic Kolsky bar PCI measurement, the timing sequence/synchronization of x-ray open-close time window (x-ray shutters), Kolsky bar firing, dynamic event, and image detection were critical, and the scheme utilized for this setup is shown in Figure 6. Since the frame separation could only be \( n \times 3.68 \mu s \), this setup took advantage of the master-clock \( P_0 \) pulses (also spaced at 3.68 \( \mu s \)) supplied by the syn-
chrotron. Using a DG535 digital delay/pulse generator (or simply DG535; by Stanford Research Systems) and the high speed camera, the $P_0$ pulses were synchronized with the $8 \times 7$ superbunch, i.e., co-center the camera shutter window ($\widehat{DE}$ in Figure 1) with the superbunch ($\widehat{BC}$) via $P_0$. The $P_0$ pulses were fed into another DG535 to skip a certain number of $P_0$ pulses (this DG535 serves as a frequency divider), and the output was then passed on to the camera as the external phase-locking signal. Upon being triggered by an external signal, the camera began to record the event.

The timing sequence is detailed as follows. At $t_{-3}$, a single-shot signal from a DG535 sent a trigger to the solenoid valve to launch the striker. A delayed trigger is then sent to the slow shutter in order for it to begin the required opening sequence which occurred at $t_{-2}$ (it completes at $t_{-1}$). Finally, the DG535 also sends another delayed trigger to the slow shutter in order for it to close at $t_1$. (Fast shutters were not used in this series of experiments.) The time window $\Delta t_{-1,1}$ defines the x-ray time window for dynamic measurements. Within this window, the strain gage on the incident bar was excited and the voltage signal triggers an oscilloscope and the camera at $t_0$, recording the dynamic event on the camera (PCI images) and on the oscilloscope (strain gage and force transducer signals; examples are shown in Figure 5). In order to set the delay to trigger to the slow shutter for a desired air pressure, the calibration curves (Figure 4) were used to determine $\Delta t_{-3,0}$; the delay setting equals to: $\Delta t_{-3,0}$ – slow shutter opening time – the jitter in $\Delta t_{-3,0}$.

9.4 Proof-of-principles experiments

The capability of the described technique is shown via experimentation on various material systems. Sand-sand particle interaction and glass impact was performed via the compression setup, while fiber-epoxy interfacial friction, ligament-bone debonding, and single-crystal silicon failure were carried out via the tension scheme. As the purpose of the study was solely to establish the use of white beam PCI measurements on materials being loaded by either a compression or tension Kolsky bar, material
systems will only be briefly outlined. In light of this, the goal of the following analysis is to provide insight into future PCI/Kolsky bar research and to verify the efficacy of the testing technique for imaging deformation processes inside of the material body.

9.4.1 Glass Compression

The glass samples used in the experiment were originally 1 mm thick microscope slides. Using a precision wet saw outfitted with a diamond rotary blade, the slides were

Figure 9.7. Representative PCI image sequences of selected materials under Kolsky tension/compression bar loading. The image size are $320 \mu m \times 128 \mu m$, $320 \mu m \times 128 \mu m$, $320 \mu m \times 128 \mu m$, $320 \mu m \times 128 \mu m$, and $320 \mu m \times 128 \mu m$ for (a)–(e), respectively. M: metal pusher; S1: sand particle 1; S2: sand particle 2; F: fiber; R: resin; b: bubble; L: ligament; B: bone.
cut into samples being 1 mm x 3 mm x 5 mm in size which were then further sanded to remove any unwanted edging. Said samples were then fitted into an aluminum housing so as to reduce transverse deformation. Steel 0.6 mm thickness gages were then cut into wedge-shaped pieces and were mounted appropriately onto the stationary quartz load cell. Likewise, sample assemblies were placed onto the end of the incident bar so as to be compressed into the steel wedge, thereby purposefully inducing a stress concentration at the wedge-tip glass interface.

Samples were then compressed at a bar velocity of 8 ms\(^{-1}\) and a representative resulting deformation history can be seen in Figure 7a. It is apparent that crack formation began near the wedge-tip and by 22 \(\mu\)s from crack visualization, the glass was completely pulverized, which is demonstrated in the final frame of the appropriate viewgraph sequence. It is important to note that in these viewgraphs, the glass sample was constrained in majority to in plane deformation via the aforementioned Al housing, thereby demonstratively showing the effectiveness of this PCI technique, which was able to penetrate through the non-transparent Al walls.

### 9.4.2 Single-Crystal Silicon Tension

Single crystal silicon micro beams were fabricated from Silicon-on-Insulating (SOI) wafers using standard micro-fabrication techniques. The beams were 4 \(\mu\)m thick, 0.6 mm in width and possessed a gage length of 4 mm. To handle specimens of this size scale, grip ends and support strips were fabricated around the test specimen in efforts to minimize any fatal beam preloading. Once the silicon beam was completely secure in the testing fixture, the support strips were removed to leave a free-standing single crystal silicon micro beam.

Beams were then pulled in tension at an incident bar velocity of 4 ms\(^{-1}\), and a representative failure viewgraph can be found in Figure 7b. As is shown, the viewgraph history clearly depicts the fineness of the fractured particle size. It is important to note that it is possible to image the failure crack history of Si micro
beams via typical light high-speed imaging, but with improvements in sample holding and size parameters along with an increase in camera frame rate, PCI could allow for the failure history within the beam thickness to be determined rather than solely surface crack analysis.

9.4.3 Sand-sand Particle Compression

The fracture of two sand grains under dynamic compression loading was performed on Ottawa sand particles being 600 μm diameter. Bulk sand was sifted and placed in an opaque FullCure720 apparatus which was printed out using an Objet Eden 350 3D printer. Steel gage pins of 0.6 mm diameter were used to crush the two sand particles housed inside the plastic body. The components were assembled using 3 mm set screws after the bored holes in the printed plastic material were tapped with proper threads.

The entire sample package was then loaded in the bar load-cell assembly and compressed at an incident bar velocity of 8 ms⁻¹, with a representative displacement/deformation history shown in Figure 7c. Unfortunately, the camera frame size was slightly too small to image the sand-sand interface upon crack propagation, but the final viewgraph does show the presence of single particle sand fracture with crack ends located at the sand-steel interface, with the crack family exhibiting a slight fan-like nature. It is important to note that the X-rays needed for PCI easily penetrated through the FullCure720 housing, allowing for radial confinement of the sand particles without loosing imaging capabilities.

9.4.4 Fiber-Epoxy Interfacial Friction

Single Dyneema SK-62 2640 dTex fibers were prepared in a fashion so as to ensure fiber de-bonding or "pull-out" at a desired location. One end of the single fiber was acceptably adhered so as to negate slip, while the other end was mounted in an insufficiently bonding epoxy system on a bed of Al foil, with the entire assembly
being placed on a cardboard placard containing a cut-out typical in single-fiber tension testing [123]. Set-screws were then adhered to the cardboard ends allowing for easy and quick mounting of the entire package into the Kolsky tension bar scheme.

Results from one of the single-fiber interfacial friction experiments can be seen in Figure 7d, when being loaded in tension by a subjecting bar velocity of 4 ms$^{-1}$. It is important to note that this experimental assembly required the impingement of several loading pulses due to inadequate striker length required by the X-ray containment hutch size. Regardless, it is of extreme interest to visualize the de-bonding phenomenon undergone by a single high-performance fiber within an epoxy matrix, as this failure process common in composite systems may allow for an extreme level of energy dissipation due to the fiber-epoxy frictional force required to slide the fiber along the interface. The ability of the PCI technique to see through the entire specimen package has a profound benefit over typical light high-speed imaging, as epoxy/resin systems are often opaque or completely non-transparent to light.

9.4.5 Ligament-Bone Debonding

Medial collateral ligaments of the knee from euthanized lab rats were gathered via careful muscle tissue removal in efforts to expose the ligament of interest. During storage and transport to the testing facility, the samples were wrapped in gauze and frozen in saline. While this is detrimental to cellular structure, previous studies has shown that the properties of collagen fibers do not change when stored in this fashion [173]. Upon testing, the knee joint was dislocated and the tibia and femur were supported in clamps mounted onto the incident bar end and load cell, respectively. The prescribed orientation allowed the tendon to be pulled directly from either bone rather than along the bone-ligament interface.

The biological assembly was then pulled in tension with a bar velocity of 4 ms$^{-1}$ and a representative ligament-bone interface history can be seen in Figure 7e. While
the failure mechanism present in this two-part assembly is still unknown, it can be said that the PCI-Kolsky bar technique may provide a superlative framework in which to image biological tissue failure phenomenon at the increased levels of strain rate common in bodily injury of soldiers and armed personnel during blast loading events.

9.5 Summary

The successful process of in situ PCI on samples being loaded by both tension and compression Kolsky bars have been described and verification of setup capabilities have been shown on various different material systems. Single crystal silicon, fiber-epoxy interfacial friction, and ligament-bone interface debonding have all been demonstrated within the Kolsky tension loading scheme while both single grain sand particle interaction and glass compression have been performed within the Kolsky compression framework. All material systems have been succinctly discussed, thereby elucidating the power of this novel testing apparatus. It has been demonstratively shown that the PCI approach surpasses traditional light governed surface imaging techniques used during high-rate Kolsky bar loading due to the effectiveness of PCI aiding in sample through-thickness failure analysis necessary to better comprehend material and structural failure processes.
10. In situ Visual Observation of Fracture Processes in Several High-Performance Fibers

Adapted from:

10.1 Abstract

Three different high-performance fibers have been imaged in situ during Kolsky bar tensile loading using two different techniques, namely optical microscopy and Phase Contrast Imaging (PCI). Kevlar® KM2, Dyneema® SK76, and S-2 Glass® fibers have been pulled using an instrumented Kolsky bar, thereby shedding light on the failure process of each fiber type. Both the Kevlar® KM2 fiber and Dyneema® SK76 fiber exhibit rupture defined by varying degrees of fibrillation, with the former typically showing longer fibrillated ends than the latter. S-2 Glass® failure was found to exhibit a brittle fracture mode at a single point, although post-mortem analysis commonly yielded disintegration of the fiber gauge length, which is concluded to occur post the initial break due to fiber snap back or bending. Finally the efficacy of utilizing the PCI technique to achieve higher levels of spatial and temporal resolution is discussed.

10.2 Introduction

High-performance fibers possess profound mechanical stiffness and strength values in their longitudinal direction and are therefore used in a vast number of engineering applications. Coupled with their low density (~1-2.4g/cm³), they are most routinely
employed in products requiring high strength and high stiffness, namely composite structures such as helicopter blades and sporting equipment, as well as in woven form such as anti-spall linings, mooring lines, and ballistic vests. Therefore, large amounts of testing have been performed on the longitudinal tensile properties of said fibers, which is generally defined by a linear-elastic stress-strain curve [21, 22, 28, 37, 132]. This is due to the extremely high level of orientation present for polymeric fibers, resulting in crystallinity values of 75-95% [22, 28]. Upon loading, these fiber types exhibit a linear-elastic stress-strain response, typically followed by a brittle fracture being defined by fibrillation for polymer fibers [144] and a point break for glass fibers [174]. Fibrillation is thought to occur due to the low transverse bonding of the molecular chains or fibrils, which is defined by Van der Waals attraction and/or hydrogen bonding [74]. It is thus highly unlikely that the ultimate strength of polymer fibers has been achieved, as failure most likely occurs due to inter-chain slippage, rather than chain scission [31, 71–73]. Indeed, current achievable strengths seen in production of Ultra High Molecular Weight Polyethylene (UHMWPE) fibers are ~3-4 GPa [140], which is only one-sixth to one-tenth of the theoretical fiber strength [28]. In order to determine the theoretical limit of an entire fiber rather than just a single molecular chain or fibril, it is necessary to ascertain an understanding of the fiber microstructure, as well as an understanding of the failure process, which can verify model predictions as well as determine if the model exhibits the proper failure mode. Therefore, an understanding of the failure phenomenon becomes of great importance, yet herein lies the main difficulty for failure analysis of high-performance fibers; rupture analysis is solely performed via post-mortem static imaging, as their size and time period of failure limit high-speed imaging visualization. The essence of the post-mortem approach is to look at both pre- and post-failed fiber fracture images via optical microscopy or SEM, and then to combine this information with a load-deformation response in efforts to conjecture a proposed failure phenomenon [37, 38]. While this is a reasonable approach for the trained eye, it still leaves room for error, especially when rupture morphologies become complex or if the material does not
exhibit a typical bulk response seen in common material testing. For example, it has been previously alluded to that aramid fibers exhibit a possible necking behavior during failure due to post-mortem rupture morphologies showing a very fine point break [22], which as will be described, is just an alternate form of the common failure mode, namely fibrillation. Furthermore, this method also becomes inadequate when examining high-strength glass fibers, which upon tensile failure, shatter into many pieces, thereby leaving the true fracture surface unrecoverable post deformation [174, 175]. To the authors’ knowledge, no work in the literature has yet shown the failure process of high-performance fiber during the actual loading procedure. Lack of this in situ analysis is inherently due to the fine nature of these high-performance fibers, which are typically only 10-30 $\mu$m in diameter. Furthermore, the failure process of ballistic fibers is dynamic in nature, thereby rendering imaging of the failure sequence during quasi-static testing quite difficult as the number of tests required to capture the said event would be extremely time prohibitive. Thus, it is the goal of this work to show the possibility of imaging high-performance fiber in situ, during a dynamic loading process. Two techniques will be discussed, namely optical light microscopy and Phase Contrast Imaging (PCI). The capability of both of these techniques will be shown via Kolsky bar tensile testing of Kevlar® KM2, Dyneema® SK76, and S-2 Glass® followed by a brief analysis of the failure process from each fiber type. Finally, improvements in temporal and spatial resolution of the two imaging techniques will also be discussed.

10.3 Experimental

10.3.1 Materials and Loading Method

Single fiber fracture studies have been performed on three different fiber types: Kevlar® KM2, Dyneema® SK76, and S-2-glass®. Fibers were extracted from the as-received yarns and then attached to aluminum mounting fixtures. Said fixtures were designed to minimize sample gauge length, aiding in the likelihood of catching
the dynamic fracture event within the active imaging window. Typical gauge lengths were roughly 200-400 μm subject to the tolerance of the sample holder. Fiber ends were carefully fixed to the metal substrates using an appropriate epoxy, with extreme efforts placed on the minimization of epoxy wicking into the sample gauge length. The mechanism used to deliver a dynamic loading to the sample was a miniature Kolsky bar apparatus, which can be seen in 10.1. As opposed to a traditional Kolsky bar, the miniature device ignores the use of a transmission bar, as the majority of the incident waveform is reflected back down the incident bar due to the large impedance mismatch between the fiber and loading bar end, rendering the traditional method of measuring the transmitted signal from strain gauges on the transmission bar impossible. Thus, for typical single fiber longitudinal tension testing, the transmission bar is replaced with a dynamic force transducer, thereby allowing for accurate determination of the sample loading history. In the case of this experimental approach, which is designed to capture the failure process of a single high-performance fiber, the load cell was not always used to detect the dynamic force history, due to space constraints of the optical assemblage, which will be explained below. Upon adequate curing of the adhesive, the sample assemblage was loaded into the miniature Kolsky bar apparatus and prior to testing, a load isolating arm was removed from the metal sample holder in order to expose the fiber specimen to free loading between the incident bar and the rigid force transducer/backstop. Using a gas firing system, the incident bar was loaded with a concentric brass striker, thereby giving the desired loading pulse with proper pulse shaping. A typical loading waveform can be seen in 10.2. Collection of the loading pulse was performed with a data acquisition unit capable of outputting a 1.5 V TTL signal, which was then used to trigger the high-speed imaging device in both of the visualization techniques described below.
Figure 10.1. Schematic of the miniature tension Kolsky bar coupled with the high rate imaging apparatus: (a) Optical light microscopy and (b) PCI.

Figure 10.2. Oscilloscope record of the incident waveform (red) and the resulting TTL output (blue) from the oscilloscope used to trigger the high-speed imaging system.
10.3.2 Optical Light Method

Fibers have been pulled in longitudinal tension on a miniature Kolsky bar outfitted with a visual light optical imaging capability as seen in 10.1a. Due to the extremely fine nature of these ballistic fibers (~10-30 μm), high-speed imaging with a traditional camera lens system was impossible. Thus, the aforementioned Kolsky bar apparatus has been outfitted with a light-microscope in order to magnify the fiber diameter to a level that is possible to collect on a CCD recording device. The total magnification of the microscope system was 20×, with the microscope lens and camera lens adaptor possessing magnifications of 2 and 10, respectively. In order to focus the camera system onto the fiber sample, the entire Kolsky bar apparatus was outfitted with a 3-axis stage, thereby allowing movement of the fiber sample with respect to the stationary imaging system (very much similar to the traditional lab optical microscopes). A high-speed Cordin 550 camera capable of taking 32 1-megapixel images was used to capture the dynamic failure event. This camera system possesses 32 stationary CCDs that are illuminated by a rotating platter possessing mirrors that direct the incoming light sequentially onto the various CCDs, thereby allowing full-frame resolution regardless of recording frame rate and magnification. Finally, in order to gain enough light to capture the dynamic failure event onto the CCDs of the high-speed camera, a high photon-flux flash bulb was used to illuminate the fiber sample. A light focusing lens system was employed to direct the flash down onto the fiber sample via a five lens system. A 2.5 in convex lens possessing a focal length of 2 in was first used to collect the light coming from the flash source. This collected light was then condensed down with a similar lens onto an objective triple lens system utilizing convex lenses of 1 in diameter and having focal lengths of 1.5 in, 0.75 in, and 0.75 in, ordered from upstream to downstream, respectively. Greatest success of this objective lens system was found when almost all flash light was directed into the end of the imaging microscope lens. As previously described, the incident waveform used to load the fiber sample was detected with strain gauges on the incident bar, and then
captured with an oscilloscope data recorder. A 1.5 V TTL signal with a 50 ns delay from the oscilloscope trigger was then output to the camera, and by utilizing the camera software, an appropriate time delay (due to the travel time of the waveform down the incident bar) was used to capture the dynamic failure event at frame rates ranging between 200,000 and 300,000 fps (frames per second). It is important to note that the fineness of these fibers inherently limits the resolution of light microscopy, as the wavelength of visual light is one-two and zero orders of magnitude less than the fiber diameter and possible fractured artifacts’ minor thicknesses, respectively. Thus utilization of this first visual technique must be implemented with caution, as ruptured surface artifacts of the fracturing process may not be resolvable.

10.3.3 Phase Contrast Imagine

Due to the deficiency in spatial resolution of the optical light microscopy method, PCI has been employed to capture the dynamic failure process of single fibers using a synchrotron X-ray source, as X-rays inherently possess wavelengths several orders of magnitude less than visible light. It has been previously shown that PCI can be used to track internal/external failure processes of materials subjected to high loading rates using a Kolsky bar apparatus [176] or gas gun system [165,177–180]. In the present work, a Kolsky bar apparatus has been integrated with the X-ray source present in beamline 32-ID at APS, which utilizes an undulator A that possesses a period and length of 2.4 m and 3.3 cm, respectively. The Kolsky bar was placed 40 m downstream of the X-ray source perpendicular to the beam path, and the X-ray beam size was controlled with adjustable slits, thereby allowing for beam size control in both the x- and y-directions. In this set of experiments, the slit opening size was 1 mm x 0.7 mm in the x- and y-directions, respectively. By altering the gap of the undulator (30-40 mm for these tests), the spectrum characteristics of the x-ray source were varied, thereby providing an increase or decrease in photon flux, thus ensuring negligible damage to the fiber mechanical properties. Verification of negligible beam damage was verified
by subsequent quasi-static tension experiments following ASTM C 1577-03 [114] on single fibers that were exposed to radiation doses similar to those of given in the PCI experiments, yielding unchanged filament elastic modulus, failure strain, and failure stress. Due to the longevity of the loading process (\(\sim 100 \mu s\)), a preliminary temporal resolution of 0.1-1 \(\mu s\) is acceptable, which defines the necessary framing rate, and ultimately the x-ray integration time for a single frame. Thus, the frame rate is limited by the pulse electron bunch separation and electron travel time around the storage ring, which inherently selects the operation mode of the APS storage ring used during the experiments. It is important to note that these experiments were performed in tandem with other material class testing, thus the run mode selected was the APS hybrid mode, which inherently limits the maximum inter-frame separation to 3.68 \(\mu s\) when using the entire superbunch for camera exposure [165]. Once X-rays were produced by the undulator, they travelled down the beamline into the hutch in which the experiment was performed. As shown in 10.1b, the sample of interest was placed into the beamline path, and during the time of loading, X-rays were allowed to impinge upon the sample. Once they passed through the specimen, they were converted into light via a single crystal Lu3Al5O12:Ce scintillator, which was placed 18 cm downstream of the sample. As the decay time of the scintillator was 45-55 ns, which was much less than the frame separation used in these experiments (minimum of 3.68 \(\mu s\)), the ghosting effect due to previous exposure was deemed negligible [165]. The light produced by the scintillator was then reflected by a 45° mirror and then magnified by a 10× lens, ultimately being recorded by a high-speed camera. The camera used for the experiments was a Photron Fastcam SA1.1 12-bit CMOS camera, capable of achieving frame rates ranging from 5,400 fps to 675,000 fps, with an inherent loss in image window size at elevated frame rates due to data transfer limitations. The camera was phase locked to an integer of the storage ring master clock, thereby providing recording windows during the x-ray passage. The camera was triggered using an external TTL (transistor-transistor logic) signal, being provided by a high-speed oscilloscope, which itself was triggered from strain gauges attached on
the surface of the Kolsky bar. Further explanation of this procedure and the timing sequence is described in Hudspeth et al. [176] and a schematic of the experimental apparatus can be seen in 10.1b. Like the optical microscopy method, the Kolsky bar was placed on a position adjustable stand, thus allowing for proper location of the sample within the beamline path. Alignment of the sample with the Kolsky bar was performed with a 4-axis stage providing x-, y-, z- displacement and rotation about the z-axis. Sample assemblies previously described were used to ensure minimal gauge length of the fiber specimens, thereby allowing for visualization of the entire gage length within the camera view window.

10.4 Results and Discussion

High-performance fibers were pulled in tension and the fracture process was imaged using both optical light microscopy and PCI via the Kolsky bar apparatus shown in 10.1a and 10.1b, respectively. Resulting loading sequences from Kevlar® KM2, Dyneema® SK76, and S-2 Glass® can be seen in 10.3 and 10.4, using the optical and PCI methods, respectively. 10.3a shows the loading history of a single Kevlar® KM2 fiber undergoing a fibrillation failure process. As can be seen from the panel sequence, fibrillation is an evolutionary process instilled with large amounts of longitudinal splitting of the fiber. This is believed to occur due to the weak transverse bonding present between the molecular chains and microfibrils, being defined by Van der Waals attraction and hydrogen bonding [74]. In contrast to this high-degree of fibrillation, 10.4a presents the failure process of a Kevlar® KM2 fiber when imaged with the PCI technique. During this loading process, the Kevlar® KM2 fiber presents a form of fibrillation, wherein both fracture surfaces break in a brittle fashion and along the axial direction, but only along one or two fracture planes, thereby resulting in a fine point break rupture morphology [132,144]. This rupture morphology has previously led to the assumption that this fiber type may undergo a necking phenomenon [22], and as will be described and further demonstrated shortly, may be an
invalid conclusion. 10.3b depicts the failure process presented by the Dyneema® SK76 fiber via imaging with optical microscopy. A low degree of fibrillation can be seen to occur during this fiber breakage, which is the typical rupture morphology presented by this fiber type. In contrast, 10.4b shows the failure process of two Dyneema® SK76 fibers, with the top filament undergoing a mass amount of fibrillation during failure. Again, as with the Kevlar® KM2 fiber, Dyneema® SK76 possesses very weak transverse bonding being defined by Van der Waals attraction between the long flexible PE molecular chains. Finally, both 10.3c and 10.4c show the failure process of S-2 Glass®, which upon inspection, demonstrates that the fiber undergoes a brittle fracture process that is located at a single point along the fiber gauge length. It has been previously thought that high-performance glass fibers undergo a simultaneous disintegration process, wherein numerous cracks propagate through the filament during failure [174,175]. These images clearly show that this failure process is not occurring for S-2 Glass®, rather the fiber is breaking in a single point, and further breakage noted in the aforementioned studies is most likely occurring during the unloading or bending of the fiber specimen post initial rupture. As the typical mode of fiber fracture analysis is performed via post-mortem static imaging, common fracture surfaces presented by each fiber type are shown in 10.5. Fracture surface imaging from both the Kevlar® KM2 fiber and the Dyneema® SK76 fiber were performed with SEM, while the rupture morphologies presented by the S-2 Glass® fiber were imaged with optical microscopy. Both low- and high-rate fracture surfaces have been imaged so as to determine if there exists any rate effects in the failure process of these three different fiber types. With regards to the Kevlar® KM2 fiber, which can be seen in the top micrographs of 10.5, this fiber type exhibits rupture morphologies of either fibrillation or axial splitting at both low and high strain-rates. As aforementioned, previous work has described the latter rupture morphology as a sort of necking behavior, as the failure surface does look to pull down to a fine breaking point as would occur during a ductile fracture process. Even though this sort of failure surface at first glance does represent a plausible necking deformation mode, it is important to
remember that the Kevlar® KM2 fiber is quite brittle, possesses a failure strain of 3-4%, and the majority of the molecular chains are oriented in the axial direction due to the manufacturing process. Indeed, as can be seen in the set of micrographs presented in 10.6, this failure process is certainly just a form of fibrillation, wherein only a single fibrillated end remains post failure. For a reference, 10.6 presents a case of the Kevlar® KM2 fiber undergoing axial splitting, which was probed with both in situ imaging and post-mortem imaging. This failure process is an example of the axial splitting phenomenon that can be deduced from the post-mortem micrograph, which is seen in the bottom right inset of the image sequence. Herein the Kevlar® KM2 fiber is undergoing a clear axial split ending in a very fine amount of fibrillation, rather than the occasionally assumed ductile fracture.

In the second set of post-mortem micrographs presented in 10.5, the typical rupture morphologies presented by the Dyneema® SK76 fiber can are shown. At both low and high strain-rate tensile failures, it is demonstrated that the fiber exhibits a fibrillation failure process, though not as drastic as that presented by the Kevlar® KM2 fiber. Rather than fine fibrillation, the Dyneema® SK76 fiber fracture surface looks to typically possess plate-like failure surfaces along with fine fibrillation [38,144]. Though not conclusive, in all micrographs found in this set of experiments, the plate-like features originated from the skin portion of the fiber, which is a reasonable failure process, as it is well known that high-performance polymeric fibers typically possess a skin-core architecture [181–184]. It is important to note that mass fine fibrillation was also uncovered in a few of the post-mortem rupture morphologies, which can be seen in the first high-rate micrograph shown in the Dyneema® SK76 section of 10.5. This mass fibrillation process was also seen using the PCI technique, which has been previously described and is exhibited by the top fiber in 10.4b.

Finally, in the third set of post-mortem micrographs presented in 10.5, rupture morphologies of S-2 Glass® are presented. It is important to note that these failure surfaces were imaged from the slight portion of the specimen which remained on each side of the sample grid, as the majority of the fiber specimen within the
Figure 10.3. High-rate optical microscopy images taken during the dynamic loading procedure, wherein: (a) Kevlar® KM2 fiber is undergoing the fibrillation phenomenon. (b) Dyneema® SK76 is undergoing a less pronounced degree of fibrillation. (c) Two S-2 Glass® fibers both fracture in a point break manner.
Figure 10.4. PCI images taken during the dynamic loading procedure, wherein: (a) Kevlar® KM2 fiber is undergoing axial splitting. (b) The top Dyneema® SK76 is undergoing massive fibrillation while the bottom SK76 fiber exhibits a less pronounced degree of fibrillation. (c) Two S-2 Glass® fibers both fracture in a point break manner.
Figure 10.5. Post-mortem micrographs of both low- and high-rate loading fracture morphologies: (a) SEM imaging of Kevlar® KM2 exhibiting both fibrillation and axial splitting at low- and high-strain rates. (b) SEM imaging of Dyneema® SK76 fibers exhibiting various levels of fibrillation at both tested strain-rates. Very often the fibrillation process depicted ribbon-like structures, more than likely originating from the skin-core structure of the fiber. (c) Optical imaging the S-2 Glass® fiber from portions of fiber remaining post loading. Note, these fracture surfaces are more than likely not the proper initial failure surfaces, as this fiber type generally fractures into many fragments due to snap back or bending post initial fracture.
Figure 10.6. In situ optical imaging of the Kevlar® KM2 fiber when loaded in high-rate tension. Clear axial splitting phenomenon, which can be deduced from the post-mortem failure surface, is shown in the bottom right inset. Upon solely viewing the post-mortem image, this failure surface mimics necking, yet as is shown in the failure sequence, this failure mode is clearly not exhibited.
gauge length was unrecoverable, which is commonly the case for high-performance glass fibers [174, 175]. Although the failure surfaces of these fiber ends clearly show a brittle fracture process, solely using postmortem failure surfaces to analyze the true fracture history of this fiber type becomes impossible as the majority of the sample gauge length is lost during the duration of the test post initial failure. Therefore, uncovering the correct failure surfaces of said fibers must be done with in situ imaging, as has been shown in the bottom image sets of 10.3 and 10.4. It can be seen from the micrographs presented in 10.3 and 10.4 that the S-2 Glass® fiber does not undergo a total disintegration process upon fracture, but breaks solely in one location. The disintegration or explosive failure phenomenon most likely occurs post initial fiber fracture, thus rendering any further fiber breakage in the now two pieces of disconnected gauge length useless in any meaningful energy transfer process. In order to ascertain a better understanding of the rupture surface within S-2 Glass® (as well as any other fine fiber type), it is apparent that higher levels of both spatial and temporal resolution are needed for a more in depth, proper analysis. Inherent in PCI is the capability of increasing the level of contrast present in the image via moving the scintillator farther away from the sample, thereby altering the defocusing distance $D$, 

$$D = \frac{(d \times l)}{(d + l)}$$  \hspace{1cm} (10.1)

wherein $d$ and $l$ represent the sample-to-detector and sample-to-scintillator distances, respectively [185]. Thus, with changes in $D$, the resolution developed in the sample is then altered, as the radius $r_F$ of a region in the sample corresponding to a point in the image is defined by,

$$r_F = \sqrt{\lambda D}$$  \hspace{1cm} (10.2)

wherein $\lambda$ represents the wavelength of the illumination source [185]. Careful attention must be paid to the location of the detection system with respect to the sample so as to achieve an adequate image resolution while still ensuring a reasonable level of contrast. With proper sample-detector positioning, camera quality thereby becomes
the limiting factor of spatial resolution in the experimental setup. The spatial resolution of the imaging technique could be increased with a decrease in pixel size or image magnification, albeit there is still a limitation of the scintillator, as it converts X-rays to light, thereby forcing light as the detected information. Other options include using a pixel array detector (PAD) that specifically detects X-rays, thus the spatial resolution limit becomes the pixel size of the PAD, or possibly projection X-ray microscopy, which utilizes Kirkpatrick-Baez optics. Of equal importance is the capability to increase the temporal resolution while imaging the failure process. As previously mentioned, the hybrid mode was used in this study, which limits the possible framing rate to $\sim 270K$ fps. Due to the extremely fine nature of the fibers of interest, relatively few x-rays with low energy are needed to illuminate the sample. Thus, alternate electron fill modes offered at APS could be used to image the failure process, such as the standard 24-bunch mode, which allows for a maximum framing rate of 6.52M fps or even possibly the 1296-bunch mode, which could allow for a theoretical framing rate of 352.2M fps [171]. Indeed, like the spatial resolution, the temporal resolution of the imaging process is limited by the camera system rather than the illumination source itself.

10.5 Conclusions

In situ imaging of the fracture process for three different archetype fiber types has been performed utilizing two different experimental approaches, namely optical light microscopy and PCI. While both techniques have yielded important information regarding the failure phenomenon from each tested material class of fiber, it is apparent that the light microscopy technique has reached a near maximum performance, and further increases in spatial resolution are limited due to the inherent wavelength of the illumination source itself. Thus, the efficacy of utilizing PCI for fiber fracture analysis has been tested, yielding promise for the approach, as large increases in spatial resolution are possible, providing that the proper electron beam run mode, camera,
and magnification lenses are utilized. Furthermore, the PCI technique more easily allows for greater temporal resolution, as each passing of an electron bunch provides the x-ray intensity necessary to sufficiently saturate the imaging detector. This is in contrast to using optical illumination, which is currently limited to the photon flux of a single flash source, thereby causing a decrease in detector saturation at increased strain rates. At the current status, basic understanding of the fracture behavior has been realized for all three fiber types. Kevlar® KM2 undergoes fibrillation, which is defined solely by the degree to which it fibrillates. The hypothesis of necking during the failure process of this Kevlar® fiber type is not seen in any of the in situ tests, rather axial splitting is detected in situ, resulting in a post-mortem fracture surface that can be easily misinterpreted as necking. The Dyneema® SK76 fiber was seen to fibrillate, and on average, the fibrillation was less pronounced than the Kevlar® KM2 fiber. Typical fibrillation was seen to include plate-like formation, most likely evolving from the skin-core structure of the fiber. Few of the Dyneema® SK76 tensile tests did exhibit mass fibrillation, but these tests were uncommon, thus should not define the typical failure phenomenon. S-2 Glass® was seen to exhibit brittle fracture at one location along the fiber length, being opposed to the previous post-mortem analysis that described the failure process as full disintegration. Further failure past the initial brittle fracture is thus concluded to occur post rupture due to fiber snap-back or bending. Most importantly, the efficacy of implementing the PCI technique for fiber fracture has been demonstrated, yielding promise in future analysis of fiber rupture with increased levels of both temporal and spatial resolution as opposed to typical high-rate light microscopy.
11. Simultaneous X-ray Diffraction and Phase Contrast Imaging for Investigating Material Deformation Mechanisms During High-Rate Loading

Adapted from:

11.1 Abstract

Using a high-speed camera and an intensified charge-coupled device (ICCD), a simultaneous X-ray imaging and diffraction technique has been developed for studying dynamic material behaviors during high-rate tensile loading. A Kolsky tension bar has been used to pull samples at 1000 s\(^{-1}\) and 5000 s\(^{-1}\) strain rates for superelastic equiatomic NiTi and 1100-O series aluminum, respectively. By altering the ICCD exposure time, temporal resolutions of 153 ns and 3.37 \(\mu\)s have been achieved in capturing the diffraction pattern of interest, thus equating to single pulse and 22 pulse X-ray exposure. Furthermore, the sample through-thickness deformation process has been simultaneously imaged via phase-contrast imaging. It is also shown that adequate signal-to-noise ratios are achieved for the detected diffraction patterns, thereby allowing sufficient information to perform white beam XRD analysis via in house software (WBXRD_GUI). Of current interest is the ability to evaluate crystal d-spacing, texture evolution, and material phase transitions, all of which will be established from experiments performed at the aforementioned elevated strain rates.
11.2 Introduction

In any high-rate loading environment, for example, penetration, blast loading, or impact, material deformation and resulting stress-states cannot be described by simple quasi-static testing parameters. It becomes evidently apparent that dynamic loading processes are inherently complex in nature, as simple rigid body motion gives way to wave propagation effects. Furthermore, in efforts to properly understand a dynamic loading condition, it is of great importance to ascertain the deformation response of each interacting material in loading regimes similar to those experienced in the event of interest, thereby giving constitutive material properties which can ultimately be used in a modeling scheme. Thus, material property characterization is routinely performed with Kolsky bar analysis or flyer plate impact. The former method loads a material under a uniaxial stress condition thereby providing stress-strain response after strain rates in the range of $10^2$-$10^4$ s$^{-1}$, but for special bar geometries, can provide strain rates as low as 10 s$^{-1}$ and as high as $10^5$ s$^{-1}$, also further depending on sample material response and size [161] [186]. The latter technique, pressure-shear plate impact, typically yields shear strain rates in the regime of $10^5$-$10^6$ s$^{-1}$ but for specific sample geometries (e.g. thin-films) has been reported as high as $10^7$ s$^{-1}$ [187]. Indeed, a wide array of strain-rates can be achieved with proper understanding and utilization of such techniques, although the experimental results provide solely an average deformation within the sample and give no understanding of localized deformation or damage propagation. Barring transparent materials, only surface tracking of damage mechanisms can be performed with high-speed optical imaging, thereby gaining a slight advantage over signal response, yet still limiting analysis onto the sample surface, leaving the interior deformation mechanisms unknown. Recently, an in-situ imaging technique has been developed wherein through-thickness damage and deformation can be tracked internally within the sample utilizing high-energy X-rays [165,176,188,189]. Employing the use of Phase Contrast Imaging (PCI), high levels of contrast can be generated for variations in material phase, and successful
through-thickness image sequences have been produced for both a variety of material types and structural interactions. Although this is of extreme interest, there is still a vast amount of information that could be generated from analyzing the X-rays which are diffracted during the dynamic loading process [190,191]. Historically, this sort of experiment has been performed on samples undergoing stepwise strain states, or at best, very low quasi-static loading conditions. In this sort of loading history, determination of crystal d-spacing, texture evolution, and phase transitions can be analyzed, but unfortunately, as was stated, is inherently limited to static stress-strain states. While this is of great interest and provides a baseline understanding of the effect of loading processes on different material or structural types, it does not allow for a richer understanding of the complex mechanisms presented during a dynamic stress sequence. For example, it has been shown that the phase change occurring during Kolsky bar loading in equiatomic NiTi displays an extreme amount of rate sensitivity [192]. The sharp transition front [193] which sweeps across the sample becomes either smeared or multiple transition zones develop, which negate the elastic-plastic-elastic stress-strain response inherent for this material when tested at quasi-static conditions, thereby yielding a mixed elastic-plastic response in the phase transition regime [192,194]. Thus, using a focused X-ray beam, a site specific phase change could be detected, thereby shedding insight into the phase change instigation mechanism occurring during dynamic loading. Furthermore, as previously stated, one could use the diffracted beam to gain insight into crystalline d-spacings and texture evolution of dynamically loaded samples, thereby aiding in the understanding of deformation mechanisms of the elastic regime in rate-sensitive materials. One could then find the strain in certain crystal directions, which can be compared to typical quasi-static loading rates, ultimately aiding in the understanding of mechanisms causing such a loading rate sensitivity. It is therefore the purpose of this work to show the possibility of performing such high strain-rate diffraction experiments on two different material types, namely aluminum 1100-O and superelastic equiatomic NiTi when being loaded in a dynamic tensile environment. A miniature tension Kolsky bar has been used to
provide loading to these materials in efforts to display the peak width change of cer-
tain crystal directions in the Al samples, and the stark phase change from austenite
to martensite that occurs for the NiTi material.

11.3 Experimental

11.3.1 X-Ray Characteristics

Kolsky bar experiments are inherently dynamic in nature, with the specimen de-
formation process generally occurring in the realm of 100-200 $\mu$s. Thus, as compared
to more rapid loading schemes such as shock wave studies developed via gas gun
experiments, Kolsky Bar loading provides comparatively ample deformation time,
thereby yielding experiments which are more easily captured in the current timing
scheme (details are described in Section 11.3.4). That said, these events are still quite
rapid, thereby enforcing the implementation of high energy, high flux X-rays which
can penetrate through the sample thickness to generate phase-contrast images and
diffraction patterns with a sufficient signal-to-noise ratio. In order to obtain such
bright X-rays, the Advanced Photon Source at Argonne National Lab has been uti-
лизed, as it produces X-rays with the temporal resolutions and energy characteristics
of consequence. In the standard run mode offered at APS, twenty-four equi-spaced
80 ps long electron bunches travel around the storage ring with a period of 153 ns,
thereby providing ample framing time for Kolsky bar loading. Furthermore, due to
the longevity of the experimental duration, multiple X-ray pulses can be used for
detection of each imaging frame via increasing the camera exposure time.

The X-ray beamline utilized for these experiments (32-ID) was produced with an
APS Undulator A, and possessed a period and length of 3.3 cm and 2.4 m, respectively.
In this case, a white beam was selected so as to increase the X-ray flux, with the
beam possessing a gaussian 2D shape, and at 10 keV, it possessed a full width at half
maximum of 0.6 mm and 1.3 mm in the horizontal and vertical directions, respectively.
Furthermore, the undulator gap is variable, providing variations in spectral flux along
with energy shifts of harmonic peaks. In the present work, the undulator gaps of interest are 20 mm and 30 mm, and resulting energy spectrum curves can be found in Figure 11.1. Compared with a monochromatic beam, the flux of a white beam is usually many orders' higher, and thereby the beam-induced heating effect needs to be carefully assessed. In a separated experiment with the undulator gap set to 30 mm, we collected single-pulse diffraction patterns from Al samples during in situ heating in a helium environment. The results reveal that the measured lattice parameters of the sample at different temperatures match the theoretical values very well. In the temperature ranging from 25 C to 300 C, the largest deviation is only 0.01%, which is much smaller than the magnitude of strain measured in our Kolsky bar experiments. When we reduced the undulator gap to 20 mm in our experiments, a single-crystal Si with 1mm thickness was placed in the upstream of beam path to filter out the low-energy photons. As shown in Figure 11.1(b), photons with the 1st harmonic energies are entirely absorbed by the Si filter. As a result, compared with the beam generated by the undulator with 30 mm gap, the integrated flux of the Si filtered beam is smaller and the harmonic energies are higher. Both of these two factors contribute to an even lower heat load on the sample. Therefore, we believe the beam-induced heating effect in our experiments is negligible.

11.3.2 Imaging and Diffraction Geometry

Figure 11.2 shows the schematic and photos of the experiment setup. In the experiments, both in-line phase contrast imaging and in-situ diffraction can be detected simultaneously. The sample orientation and camera geometry have been utilized and prepared so as to allow for transmission diffraction, as the reflection mode requires greater amounts of bar rotation with respect to the beamline, and inherently hinders small angle detection due to the placement of the backstop and mounted load cell. The phase-contrast images of the samples were collected by an optically coupled high speed camera (Photron FastCam SA1.1), while the diffraction patterns were recorded
Figure 11.1. X-ray energy spectrum for the (a) aluminium and (b) NiTi specimens. The aluminium tests used an undulator gap of 30 mm, and the first harmonics become of most interest. The NiTi samples used an undulator gap of 20 mm, which yields X-rays with much higher flux. In efforts to mitigate this sample absorption effect, a single-crystal Si wafer (1 mm thickness) was placed upstream of the X-ray path to filter out the low-energy photons.
using an optically coupled intensified charge-coupled device (ICCD, Princeton Instruments PI-MAX). The scintillators for imaging and diffraction detection are LuAG:Ce (100 \( \mu m \) thickness) and LYSO (300 \( \mu m \) thickness) crystals, respectively. The ICCD was mounted into a rotation arm, which allowed for the camera to be continually located in the typical circular track configuration as compared to an arbitrary position. The circular track configuration could facilitate ease of pre- and post-analysis of white-beam diffraction patterns and the control of detection angle. A Huber 410 Goniometer was used to control the rotation arm with an angle resolution of 0.001°. It was possible to adjust the sample-to-detector distance by sliding the camera along a linear rail on the rotation arm, and a picture of the assembly has been included in Figure 11.2(b).

The diffraction work performed in this study was aided via the in-house program WBXRD.GUI. This Matlab-coded software has been developed for simulating and analyzing white-beam diffraction patterns from polycrystalline samples. It is particularly useful when one deals with (single-pulse) noisy diffraction patterns, and an area detector is placed with an offset angle from the incident direction, i.e. the detection plane is not perpendicular to the incident beam while the transmission spot may not fall on the detector, as shown in Figure 11.3(a). For a given detector location and X-ray energy, the scattering vector \( q \) and azimuthal angle \( \varphi \) at each pixel position on the detector can be calculated. Figs. 3(b) and 3(c) show examples of \( q \) and \( \varphi \) maps for a detector angle of 25° and X-ray energy of 12.9 keV (i.e. 30 mm undulator gap) assuming the beam position is vertically centered with the detector.

The simulation of a white-beam diffraction pattern from a known material starts from the calculation of monochromatic beam diffraction patterns for the specific detector location, \( I(\theta, E) \). Then these mono-beam diffraction patterns are integrated over the entire energy range with the weighting factor being the flux of photons with different energy, \( F(E) \). Here, the intensity profile is described using the pseudo-Voigt function. The sample crystal structure and the diffraction intensities \( I \) of different atomic planes (\( hkl \)) are input parameters, which can be obtained from Interna-
Figure 11.2. (a) Schematic of the integrated X-ray system and Kolsky bar apparatus. Note that the standard 24-bunch mode offered by APS was utilized. In the current configuration, both transmission X-ray diffraction and PCI were detected simultaneously during the high-rate Kolsky bar loading via an ICCD and high-speed camera, respectively. (b) Photographs of the setup within the 32-ID-B hutch. Demarcations have been placed on vital pieces of equipment in both images including: (1) high-speed camera used for PCI; (2) miniature tension Kolsky bar; (3) helium-filled X-ray flight path; (4) ICCD camera used to record the diffraction pattern during high-rate loading; (5) rotating camera arm; (6) goniometer; and (7) positioning stage used to mount the fast-response load cell. Note that there is a compression Kolsky bar also located on the bar frame, thereby allowing for either tension or compression testing depending on bar selection.
Figure 11.3. (a) Diffraction geometry. $O$ is where the X-ray beam and the sample interact; $O'$ is the transmitted beam position on the detector plane; $A$ is the scattered beam position on the detector; $k_i$ and $k_f$ stand for the wavevectors for the incident and out-going beam, and $q$ is the scattering wavevector. $2\theta$ (i.e. angle $O'O_A$) is the diffraction angle. (b, c) The pixelated scattering vector $q$ map (b) and azimuthal angle $\varphi$ map (c) on the ICCD, when the detector rotation angle is 25°, X-ray energy is 12.9 keV and the beam position is assumed to be vertically centered with the detector.
tional Center for Diffraction Data (i.e. JCPDS cards). For each x-ray energy $E$, the WBXRD.GUI calculates the reciprocal lattices and projects diffraction information onto the detector plane. These mono-beam diffraction patterns are then integrated over the entire energy range with the weighting factor being the flux of photons with different energy, $F(E)$. Here, $F(E)$ reflects the post-sample flux density, which has been modified by considering the sample absorption of photons with different energies.

$$I_{\text{white}} = \int_{E_1}^{E_2} I(\theta, E) F(E) dE \quad (11.1)$$

For our case, $E_1$ and $E_2$ are typically 1 keV and 60 keV, respectively. Photons with energy higher than 60 keV have much lower flux (Figure 11.1) and the corresponding diffraction scattering angles are very small, thereby their diffraction contribution is not considered. To improve the calculation speed, discrete diffraction peaks are considered, meaning that $I(\theta, E)$ is normally replaced by a series of $I_{hkl}(\theta, E)$. Equation 11.1 then becomes

$$I_{\text{white}} = \int_{E_1}^{E_2} \left[ \sum_{i=1}^{n} I_{hkl}(\theta, E) \right] F(E) dE \quad (11.2)$$

Essentially, the white-beam diffraction intensity at a given scattering angle is the convolution of the input diffraction intensity for different atomic planes with the energy spectrum of the x-rays. Note that the polarization of x-ray is not considered in the simulation.

### 11.3.3 Kolsky Bar Apparatus

In order to provide a dynamic tension loading environment, a modified Kolsky bar apparatus was employed. The bar material was aluminum and possessed a length and diameter measuring 225 cm and 12.7 mm, respectively. A brass striker tube, running concentric with the incident bar, was fired from a teflon housing by the release of compressed air via solenoid actuation. The striker then impacted a flange located
on the incident bar end, thereby generating an elastic tensile stress pulse. The usual travel time between the solenoid firing signal and striker-bar impact was \( \sim 120 \) ms and possessed no more than a 10 ms jitter. Typical striking velocities ranged from 3-5 m/s. Post impact, the elastic stress wave was recorded via two strain gauges located on the incident bar surface, being diametrically opposed so as to minimize temperature and bending effects. As the stress wave continued down the bar length and impinged upon the sample, the majority of the incident waveform was reflected as a compression wave, which can be verified in Figure 11.4(a). Location of the strain gauges was specifically designed to ensure that overlap in the incident and reflected pulses would be non-existent and also to ensure a necessary time window elapsed for shutter activation, which is described in Section 11.3.4. The strain gauge signal was passed through a Wheatstone bridge, powered by an excitation voltage from an Agilent E3630A DC power supply, and the resulting output voltage was amplified with a Tektronix ADA 400A Differential Preamplifier. In this experimental configuration, due to constraints driven by hutch size limitations, the traditional transmission bar was replaced with a Kistler 9712B5 fast-response quartz based load cell, which is acceptable if the impedance mismatch between the bar end and sample is large. The resulting output signal was then sensed and amplified by a Kistler 5010 dual mode amplifier. Both the amplified signal generated from the incident bar strain gauges and the transmission load signal were collected with a two-channel Tektronix DPO 4032 Oscilloscope. Further description of the loading scheme has been described in previous works [176,189]. In order to ascertain the strain rate (\( \dot{\varepsilon} \)) and strain (\( \varepsilon \)) incurred by the dynamically loaded sample, the well known Kolsky bar relations have been utilized and are described in Equations 11.1 and 11.2,

\[
\dot{\varepsilon} = \frac{C_b}{l_s}(\varepsilon_I - \varepsilon_R) \quad (11.3)
\]

\[
\varepsilon = \int_0^t \dot{\varepsilon} d\tau = \int_0^t \frac{C_b}{l_s}(\varepsilon_I - \varepsilon_R) d\tau \quad (11.4)
\]
wherein $c_b$, $l_s$, $\varepsilon_I$, $\varepsilon_R$, represent the incident bar acoustic wave speed, sample gauge length, incident signal bar strain, and reflected signal bar strain, respectively. As previously stated, due to hutch size limitations and the specifically designed large impedance mismatch between the sample and incident bar end, the typical Kolsky transmission bar has been replaced with a fast response load cell, thereby allowing for direct detection of sample axial force ($F$). This force is divided by the sample initial cross-sectional area ($A_o$) in order to obtain the average instantaneous longitudinal engineering stress ($\sigma = F/A_o$) throughout the duration of loading.

In order to deliver a valid stress-strain response from a loaded material, it is imperative that the strain rate experienced during the experiment be of a constant value. Thus, typical strategic pulse shaping is employed which allows for modification of the incident stress wave shape, ultimately being tailored to meet the response of the tested material with an educated trial and error approach [195, 196] or via analytical solution [197]. The former, which is more often used due to its ease, has been sufficiently performed on NiTi samples similar in nature to the currently tested specimens, but was performed in a compression environment [192, 194]. In the current experimental design, the cross-sectional area of both the Al and NiTi samples were of such insignificant size with respect to the incident bar, that the vast majority of the incident signal is reflected due to the extremely high impedance mismatch. In this case, it is then desired to generate a trapezoidal pulse, and only slight pulse shaping is needed to reduce the high-frequency stress response. A typical set of recorded waveforms can be seen in Figure 11.4(a), which demonstrate the trapezoidal shape of the incident and reflected waveforms. Achievement of a constant strain-rate is corroborated by Figure 11.4(b), which shows the strain-rate history of the loaded sample along with the specimen stress recorded by the fast-response load cell.
Figure 11.4. (a) Raw voltage signals relevant to the high-rate loading produced from the Kolsky bar setup. Note that in lieu of a transmission bar, a fast-response quartz load cell has been utilized, which results in a reduced experimental framework footprint required by the constrained hutch dimensions. This load response is represented by the dotted black line. Furthermore, this load detection approach is viable if the impedance mismatch between the incident bar end and sample is quite large, which is demonstrated by the similarity between the incident and reflected waveforms shown by the solid black line. (b) Resulting force and strain-rate histories represented by dotted black and solid black curves, respectively. Note that the sample is loaded into the constant strain-rate regime.
11.3.4 Timing Sequence

In order to protect the scintillator for PCI and reduce unnecessary heat load on the sample, temporal bracketing of the experimental period of interest becomes necessary and is therefore described in detail. Effective timing for experimental delays was achieved via careful development and execution of a double window timing scheme composed of a set of fast shutters and slow shutters. The slow shutters consist of water-cooled bulky copper blocks which can bear the heat load caused by the intense white beam, while the fast shutters are made of small rotating lead blades that open and close much faster than the slow shutters. A schematic of the entire sequence can be seen in Figure 11.5. The external window, referenced from $t_{-3}$ until $t_4$, controls the slow shutter system and is governed by delays referenced from $t_{-4}$, namely firing the solenoid, which releases gas into the gun barrel thereby accelerating the striker. This external timing window is designed to possess a slow shutter full opening time of roughly 40 ms ($t_{-2}$ to $t_3$), which can adequately account for the striker to incident bar impact time variation. Inside of the slow shutter open time window lies the internal timing window operating a pair of fast shutters, which further bracket the actual experimental time of interest down to a few milliseconds ($t_{-1}$ to $t_1$). Triggering for the PCI camera was sent immediately at $t_0$, as the onboard memory storage allowed for ample total time of recording, being much longer than the duration of the entire experimental event. In contrast, triggering of the ICCD camera required special attention, as in the case of this experiment, the system was only able to capture one single frame. A system of sequential delay changes was thus instilled over multiple experimental events so as to build up a diffraction pattern evolution during the entire loading history. As the purpose of this set of experiments was to show the efficacy of achieving a necessary signal-to-noise ratio during small time windows throughout the entire Kolsky bar loading event, this capture method was reasonable, but upon future work, installation of a multi-frame intensified imaging system would be more fruitful, thereby allowing for the capture of a diffraction pattern evolution during
an entire loading sequence. Thus, for each dynamic tensile experiment, a multi-frame PCI sequence was recorded with a high-speed camera, along with a single diffraction pattern that was registered onto an ICCD within the duration of loading. Furthermore, as is described in Section 11.4.1, single pulse diffraction was possible for the aluminum samples tested and eight pulse diffraction was achieved for the NiTi samples, thereby resulting in a possible framing rate of 6.5M fps and 815K fps, respectively.

11.4 Results and Discussion

11.4.1 Aluminum

Inclusion of aluminum 1100-O into this experimental sequence as the first trial material was two-fold. First, the material response of Al has been well characterized via mechanical loading and thus provides a reasonably well understood stress-strain response at high loading rates [198,199]. Second, diffraction analysis of Al embodies a large sect of research, and thus a plethora of diffraction analysis already exists [200]. Ultimately, the goal of this material class is to show the efficacy of ascertaining strains in specific crystal directions via peak broadening.

In order to generate a high-rate tension loading environment, the miniature Kolsky bar apparatus described in Section 11.3.3 was utilized, thereby generating a strain rate in the regime of 5000 s$^{-1}$. An average resulting stress-strain curve is plotted in Figure 11.6, showing agreement with previous work [198,199]. On the said stress-strain figure, symbols have been plotted which demarcate the strain at which a diffraction pattern was taken. In the current setup, only one pattern was capable of being recorded during the short-duration of interest, being 3.37 $\mu$s duration for the Al tests, therefore relating to an integration of 22 X-ray pulses. This duration has been chosen to maximize signal intensity, thereby increasing the signal-to-noise ratio and thus allowing for valid diffraction pattern analysis. For a reference, a raw-unprocessed 65-pulse Al diffraction pattern has been included and can be seen in Figure 11.7(b). It
Figure 11.5. Schematic of the timing sequence used throughout the experimental duration. At time $t_4$, a delay generator (DG) signal is sent to open the gas gun valve, thereby firing the striker. Governed by a specified delay dictated from a predetermined striker travel time, an additional DG signal is sent to open the slow shutter at time $t_3$, which is the opening bracket for the outer experimental window. Note that the slow shutter opening signal is sent early enough to ensure that the shutter window is completely open before the striker impacts the incident bar end. Upon striker impact, a tensile stress wave is sent down the incident bar and, as this wave passes through the bar at time $t_0$, it is detected by a set of strain gauges, which are located 84 cm from the sample interface. Upon detection of the stress wave, a delayed DG signal is sent to close the fast shutter system, demarcated by time $t_1$. Finally, an ultimate DG signal is sent to close the slow shutter system at time $t_3$, thus closing the outer experimental time window.
Figure 11.6. Representative stress-strain curve from the 1100-O aluminium samples pulled in tension. The current data, representing a testing strain-rate of $5000 \text{s}^{-1}$, is demarcated with the solid black curve, and for comparison previous data performed at $1000 \text{s}^{-1}$ [198,199] have also been included. The various symbols which are overlaid on the plot dictate stress-strain states at which a diffraction pattern has been recorded with the ICCD. In-depth analysis from only an unloaded sample and a sample loaded up to demarcation 10 have been included in the in-depth diffraction pattern analysis shown in Figure 11.8, as the current goal of this study is to show the possibility of performing such high-strain-rate loading while simultaneously capturing high-frame-rate diffraction patterns and phase-contrast images.
Figure 11.7. Experimental diffraction patterns wherein the ICCD gating time was varied so as to allow for (a) single-pulse diffraction and (b) 65-pulse diffraction. Azimuthal integration of each pattern was performed to display the corresponding one-dimensional intensity plots for (c) single-pulse diffraction and (d) 65-pulse diffraction. Clearly the multi-pulse gating yields a much higher signal-to-noise ratio, but note that there is still enough information in the single-pulse image to detect a distinct ring pattern and well defined peak position via post-processing.

is important to note that single pulse diffraction pattern capture was possible for the Al samples, and a resulting pattern has also been included in Figure 11.7(a), both patterns being generated from 225 μm thick samples. Furthermore, integration of both patterns was performed to display the corresponding 1D intensity plots for single pulse and 65-pulse diffraction and the results are displayed in Figures 11.7(c) and 11.7(d), respectively. Pulse duration parameters have also been succinctly included in Table 11.1.
Figure 11.8(a) and 11.8(b) show the diffraction patterns from an Al sample collected before and 30 μsec after the start of the tensile pulling, respectively. These patterns were formed by 22 pulses of X-rays with the energy spectrum shown in Figure 11.1(a). Note the intense ring-shape feature at corners of each pattern is the edge of the scintillator. In both patterns, two diffraction peaks can be observed. The peak at lower angle (left) is attributed to Al (111), and the peak at higher angle (right) is Al (200). Both are generated by X-rays with the 1st harmonic energy (12.94 keV). The 2nd harmonic (311) and (222) peaks are present between these two major peaks. However, as the flux of X-rays with the 2nd harmonic energy is less than 3% of that of the 1st harmonic X-rays, the (311) and (222) peaks are overwhelmed by the intense 1st harmonic (111) peak. Figure 11.8(c) depicts the 1D diffraction intensity profiles (open symbols) and corresponding theoretical simulations (lines) of unstrained and strained states of the Al sample. The 1D data was obtained by integrating the 2D patterns over the azimuthal angles ranging from 173° to 187°, as indicated in Figures 11.8(a) and 11.8(b). In the simulations an anisotropy parameter (i.e. \( I_{111}/I_{200} \)) was considered to account the initial texture structure of the Al sample. Figure 11.8(d) shows a closer look of the (111) peaks, and a shift of the peak position to the lower angle due to the pulling can be clearly observed. The simulation indicates a 0.25% lattice expansion of the Al sample along the tensile loading direction. The quantitative agreement between the data and simulation demonstrates that ultrafast white-beam diffraction is capable of measuring elastic strain with extremely small magnitude. Thus, rather than solely determining an average strain measure over the entire sample during Kolsky bar loading, it is now possible to additionally analyze elastic strain in a specific loading direction at extremely high temporal resolutions.

11.4.2 NiTi

Inclusion of the NiTi super elastic material class was chosen due to the stark phase-change which occurs during the austenite to martensite phase transition, being
Figure 11.8. Diffraction patterns from the Al sample, collected (a) before and (b) 30 ms after the start of tensile pulling. (c) Diffraction intensity and corresponding simulations from unstrained and strained Al samples. The one-dimensional data profiles were obtained by radially averaging the two-dimensional diffraction patterns shown in (a, b) from azimuthal angle 173° to 187°, as indicated. (d) A closer look at the (111) peaks, showing the shift of the peak due to a lattice strain of 0.25%. Note the disparity in elastic strain of the (111) peak as compared with the large amount of average plastic tensile strain demonstrated in Figure 11.6.

Table 11.1. Experimental parameters for each of the chosen conditions. Note the high level of temporal resolution.

<table>
<thead>
<tr>
<th></th>
<th>Aluminum</th>
<th>NiTi</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>Short Exposure</td>
<td>Long Exposure</td>
</tr>
<tr>
<td>Undulator gap (mm)</td>
<td>30</td>
<td>30</td>
</tr>
<tr>
<td>Detector angle (°)</td>
<td>25</td>
<td>25</td>
</tr>
<tr>
<td>Shutter time (μs)</td>
<td>0.153</td>
<td>3.37</td>
</tr>
<tr>
<td>Temporal Resolution (ns)</td>
<td>0.1</td>
<td>3370</td>
</tr>
<tr>
<td>Number of X-ray pulses</td>
<td>1</td>
<td>22</td>
</tr>
</tbody>
</table>
typically described as stress-induced-martensite (SIM). Furthermore, understanding of the dynamic loading process in NiTi alloys is of extreme interest, due to their usage in medical applications, stress-mitigation structures, and energy absorbing devices. It is well known that the SIM transformation is quite rate-sensitive which results in an alteration of the elastic stress plateau [192,201,202].

Similar to the Al samples, deformation of NiTi has been impinged via Kolsky bar loading, resulting in a strain-rate of $\sim 1000 \text{ s}^{-1}$, and the stress-strain response has been recorded and is displayed in Figure 11.9. Only single diffraction patterns could be collected for one loading duration, thus rendering the use of multiple loading sequences necessary in order to build up a representative response from the deformation history. Each of the aforementioned demarcations relates to a specific stress-strain location wherein a diffraction pattern was recorded. In this set of experiments, the detector collection time was $3.37 \mu\text{s}$, thereby resulting in a collection of 22 X-ray pulses, which is deemed reasonable as the duration of loading is 100-200 $\mu\text{s}$. The undulator gap was set to 20 mm and the corresponding X-ray energy spectrum is shown in Figure 11.1(b).

Figures 11.10(a) and 11.10(a) show the diffraction patterns from a NiTi sample, collected before loading and 1.75 msec from the start of loading, respectively. Twenty-two X-ray pulses, with the energy spectrum shown in Figure 11.1(b), were used to generate these patterns, which clearly indicate that the NiTi sample experiences a phase transformation during the dynamic tensile deformation. Radially averaged 1D diffraction profiles of these two states of NiTi are shown in Figure 11.10(c), and they can be well indexed as austenite and martensite phases, respectively. The different colors of the indexing bars represent different harmonic energies. The solid bars mark the peak positions of the austenite phase, while the dashed bars mark those of the martensite phase. In the diffraction data obtained during the dynamic loading, the small peak around $21^\circ$ can be observed, as pointed by the arrow, which is attributed to the austenite phase. The presence of this peak indicates the incomplete phase transformation of NiTi at the present situation [203]. This example of NiTi shown
Figure 11.9. Representative stress-strain curve produced from the equiatomic NiTi super-elastic pulled in tension. Tests were performed at 1000 s\(^{-1}\) and are transformation demonstrated by the solid black curve, which is compared with the dotted black line, representing previous data performed at 1200 s\(^{-1}\) [202]. Similar to the aluminium tests, symbols have been overlaid on the plot demarcating stress-strain states at which diffraction patterns have been captured with the ICCD.
here demonstrates the capability of the ultrafast diffraction technique in probing the transient phenomena in hard crystalline materials, such as rapid phase transformation.

Figures 11.11 and 11.12 are presented to further demonstrate the simultaneous imaging-diffraction nature of these experiments. The former image sequence depicts a set of phase contrast images derived from a single dynamically loaded NiTi sample being loaded at a $\sim1000 \text{ s}^{-1}$ strain-rate, while the latter illustrates single diffraction patterns captured from various samples that have been loaded via similar conditions. Each image in Figure 11.12 thus presents a diffraction pattern which corresponds to a similar stress-state at which the phase contrast image presented in Figure 11.11 was collected. While the real-space images could only show limited structural information of the NiTi sample during high-rate loading, the diffraction patterns clearly reveal the phase transformation process. As shown in Figure 11.12, the NiTi transforms from the original austenite phase to the martensite phase (from 3rd frame to 4th frame) upon tensile strain, and changed back to the austenite phase after the sample fractured and the strain was released in the end (from 6th frame to 7th frame). As captured by the current data set, NiTi clearly undergoes the SIM transformation during dynamic tensile loading in the $1000 \text{ s}^{-1}$ strain-rate regime being concretely verified by the slight pattern shift shown in Figures 11.10 and 11.12. As shown, the ambient cubic crystal structure reorients to the martensitic phase exhibiting a monoclinic structure during loading and then upon failure, returns to the initial cubic structure.

Furthermore, the NiTi example shown here is intended to demonstrate that the ultrafast diffraction can well complement the established PCI technique, and help provide additional insight into the material deformation process. Also, it underscores the need of proper analysis tools in understanding the complex white-beam diffraction data. Although the present work required multiple experimental runs to build up a representative diffraction pattern evolution, it can be clearly seen that it is possible to gather fine temporal resolution pattern evolution during the high-rate loading sequence.
Figure 11.10. Diffraction patterns from the NiTi sample, collected (a) before and (b) 1.75 ms after the start of tensile pulling. (c) One-dimensional diffraction intensity profiles, obtained by radially averaging the two-dimensional patterns with all available azimuthal angles. Reference peak positions of the austenite (solid lines) and martensite (dashed lines) phases are displayed at the bottom, and the blue, red, green and cyan colors represent peaks that correspond to the second, third, fourth and fifth harmonic energies, respectively. Note that the photons with the first-harmonic energy have been entirely absorbed by the Si filter.
Figure 11.11. PCI sequence captured with the high-speed camera showing the deformation process of the NiTi material throughout the loading process. Frame times have been chosen to correlate with the stress-strain states at which diffraction patterns displayed in Figure 11.12 were captured. Note that this image sequence consists of snapshots within one loading event.

Figure 11.12. Diffraction pattern sequence captured with the ICCD which shows the evolution of the martensitic transformation at sequential stress states during the dynamic loading process. It is important to note that each pattern is captured from a different sample within the tensile loading history at stress-strain states represented by the appropriate symbols in Figure 11.9 and at delay times shown in Figure 11.11. The pattern taken at $t = 1.75$ ms was captured within the PCI image sequence shown in Figure 11.10.
11.5 Conclusion

In order to better understand the deformation process of materials loaded in a dynamic environment, both equiatomic NiTi and Aluminum samples have been pulled in tension with a miniature Kolsky bar, while simultaneously performing X-ray PCI and diffraction. Using synchrotron radiation generated at beamline 32-ID-B of the Advanced Photon Source at Argonne National Lab, high-temporal resolution X-ray diffraction images can be generated, which thus allow for legitimate material analysis of dynamically loaded samples. Evolution of sample texture, elastic crystal straining, and material phase can all be analyzed using in-house software (WBXRD_GUI), with the latter two being immediately confirmed in this work. Furthermore, with the described experiments and data analysis, it has been shown that white-beam diffraction is sufficient to perform the aforementioned investigation. While the majority of this work displays a diffraction pattern temporal resolution of 3.37 $\mu$s, it has also been shown that a resolution of 100 ps is possible using the standard 24-bunch mode offered at APS.
12. Summary

Although all internal chapters have terminated with a brief overview of the specific topics covered within each respective section, it is deemed of use to provide a few short and concise concluding remarks, specifically with regards to the final efforts made in both Volumes 1 and 2, along with immediate future work proposed for each. As such, both volumes will be discussed separately.

The first, and primary objective of this work has been to better understand the means of failure in high performance fabric. Although a myriad of energy dissipation mechanisms and governing characteristics exist below the ballistic limit of the system, specific emphasis has been directed at velocities causing local failure in the fabric system. In other words, velocities that have caused part, if not all, of the fabric to exhibit a rapid local failure, thereby dissipating minimal energy from the impacting projectile. These sort of impact velocities surround the ballistic limit of a multi-ply fabric system. In order to assess the local rupture of constituent yarns within a fabric system subjected to transverse impact, several experiments have been performed on the single filament and single yarn level efforts to attach an event that is much more tractable. Specifically, great interest has been paid to the single yarn transverse impact experiment that has been used somewhat sparsely throughout high-performance fabric literature, due to its rather odd experimental setup and closer to fundamental science nature, as compared to full system impact. Additionally, it inherently requires access to a high speed imaging device. As with full fabric impact, at sufficiently high impacting velocities, single yarn also exhibits a transition from long range deformation to rapid, localized failure upon projectile-yarn contact. Interestingly, analytical and numerical prediction of this ‘critical’ velocity over-predicts the velocity at which failure occurs immediately in actual experiments, by nearly 50%. Such a stark over-prediction is really what pushed this piece of work forwards in efforts to
understand how filaments actually fail during transverse impact, and what should in reality be the proper failure criterion. Initial efforts to understand this localized failure were placed on strain rate sensitivity and the effect of transverse compression, but little detrimental effect was found for either. Interestingly, it was found that throughout literature, nearly every single piece of work regarding high-performance fabric assumed that the failure of the filament occurred as pure tension, regardless of loading condition. Standing on the periphery of historical fabric literature, this appears to be somewhat of an odd assumption, but due to the extremely fine nature of these filaments (∼10 μm in diameter) it was believed that no impacting surface nor any loading geometry could possibly promote local bending or shearing stresses that could possibly overcome such a fine diameter. Thus, effort has been directly placed at attacking this assumption, and has been done so in a variety of ways throughout this text. Firstly, filaments were placed in a torsional shear stress environment, which demonstratively showed a reduction in tensile failure strain with increasing levels of shear stress. Although not the same geometry that is developed during transverse impact, the proof that an alternative stress state could indeed alter the failure stress of a single filament spurred on further efforts directed at recreating the geometry that is developed during an actual transverse impact environment. Subsequently, a quasi-static loading environment was developed with an emphasis placed on assessing both the angle of loading (similar to the ‘tent’ formation during yarn transverse impact) and the radius of curvature of the indenter. It was determined that loading angle and projectile geometry clearly affected the resulting failure strain of the fiber. To the authors knowledge, such a finding is the first study demonstrating that a filament in a transverse impact loading geometry does is indeed affected by the loading conditions. This was then immediately followed by actual yarn transverse impact experiments, where a large effort was focused on determining the yarn transverse impact critical velocity using various projectile nose geometries. Again, as with the quasi-static experiments, it was found that projectile nose geometry is a clear governing parameter of the failure of the high-performance fiber; this was further corroborated by micro-
graphs depicting large changes in fracture surfaces when impacting with the different nose geometries. Said experiments were also used to determine the longitudinal wave, transverse wave speed, and load developed behind the longitudinal wave speed, all as function of the nose geometry and strike velocity of the impacting projectile. Finally, a good deal of effort was placed on using high speed imaging to look closely at the impact zone from various angles, in efforts to determine the means of yarn failure during the transverse impact event. After all such experiments, it can be said with confidence, yarns do not fail in pure tension during transverse impact and as such, models must reconsider their implemented failure criterion. As for immediate future work it is suggested that effort be placed on understanding the actual stress state developed in a filament/yarn directly around the projectile head. Although effort has been made in the current work with the use of Finite Elements Analysis, it is suggested that a much more mature modeler focus his/her effort on understanding the deformation process around the projectile head, as the current author does not possess the required level skill, expertise, or time to pursue this important endeavor. That said, efforts in the current work have placed in the current work on providing the required model inputs and desired outputs such as boundary conditions, loading conditions, and measurement of output physical response.

The second main topic of this thesis has been on the development and use of high-frequency X-rays in order to illuminate in-situ, internal damage mechanisms exhibited by various materials during high-rate loading. As mentioned in the opening of this document this portion of the work was highly collaborative, and in no way shape or form would any of this work have been possible without the help of numerous graduate students and and staff scientists present at the Advanced Photon Source (APS) of Argonne National Lab (ANL). Again, a high level of thanks is offered up to anyone and everyone involved in this work. Although the current work only discusses the development of two techniques, namely high-rate X-ray imaging and high-rate X-ray diffraction during dynamic loading, both of these techniques have been used on a variety of material classes and for more depth material analysis and
characterization. Although topical findings were presented for a few main material classes (metal, ceramic, polymer), the goal of both of these techniques in the scope of this document, was to perform both imaging and diffraction on high-performance polymers. While the former is shown in this text, efforts are ongoing with regards to the latter topic and will hopefully be subject of future report. With regards to recommendations on future work, this is a bit hard to direct, as this technique is extremely powerful and can be used on many different material classes in a variety of different loading geometries. It can be stated with near certainty that future efforts Professor Chen’s group will involve the use of a light-gas gun, which is in its final stages of construction, and should hopefully be integrated into the beam line of APS during the upcoming summer of 2016.
13. Recommendations

Although projecting one’s own thought on the direction of future research is albeit, a bit imperious, especially for a rather inexperienced researcher, it is deemed somewhat reasonable, even necessary, as any aid that can be provided to a topic that has an affect on the lives of future soldiers is of paramount importance. That said, it must be restated that the present author barely has justification in providing ‘expert’ guidance in this sole subject area, so caution is advised; the PhD process in it’s definition bores down the thoughts and efforts of the captive on to one sole, menial slice of a research field. Take the few following suggestions with at least two grains of salt.

First and foremost, it is suggested that most important research direction, and the topic of immediate future research efforts following this study, should be ballistic limit experiments on both Kevlar® and KM2 Dyneema® SK76 fabric, possessing identical filament denier to those found throughout this document (it may be possible use previously existing KM2 data that is available throughout literature, but caution should be taken in such data gathering, as there appear to be various forms of KM2 which is, or at least has been, produced by by DuPont™). Reasoning for the selection of these two specific filament types is two-fold. First, they represent the most common form of aramid (Kevlar® KM2) and UHMWPE (Dyneema® SK76), and are used routinely in ballistic protection environments. Second, within the currently described data set, both Kevlar® KM2 and Dyneema® SK76 have been subjected to longitudinal tension (both single filament and full yarn), torsion (single filament), transverse off-axis loading (single filament), and single yarn transverse impact. Additionally, efforts from the Chen lab have also been directed at transverse compression loading of single filaments [204], and both Kevlar® KM2 and Dyneema® SK76 have been thoroughly analyzed. In short, a battery of experiments have been performed on these materials, all aimed at understanding constituent behavior within the full fabric; now it is time
to push this understanding up another level via comparison with full fabric experiments. Of course such experiments are highly cost prohibitive, and do require a good deal of time and experimental capability, but such is the life of a diligent graduate student. If a proper sponsor can allocate the appropriate funds to ascertain fabric and projectiles, such a battery of experiments should be well within the reach of a heavy Masters thesis or a large portion of a PhD study. It is suggested that these fabric experiments be performed with three different projectile types, namely sphere, right circular cylinder (RCC), and a sharpened cylindrical punch. Although several different projectile masses could be probed, such experiments would drastically increase the number of experiments for only slightly increased fabric performance understanding. Finally, different fabric thicknesses must be tested, potentially 4 or 5 different thicknesses for both material types.

Performing transverse impact experiments on full fabric, one could then use previous work on these two material types to dig deeper into the understanding of both the Smith solution and the Cunniff equation, hopefully making a connection between the two which could ultimately lead to a much cheaper experimental testing approach in determining full fabric performance capabilities. It is also suggested that one could provide modification to the Cunniff parameter, thereby allowing for better prediction to fabric performance when using different material classes. Such a modification would come directly on the heels of understanding of the modifications made to the Smith solution, as both seem to be inherently governed by the same physical phenomenon. Inherently, successful modifications to either will provide the understanding to make changes to the other.

Along the lines of ballistic testing, it is also of interest to perform transverse impact into condensed packs composed of HB10 tape, which was tested on the diffraction experiments described in Volume II. Such a material is of great interest to future military protection systems due to the inherent uni-direction (UD) nature of the material, while removing the presence of the required epoxy/resin that is found in a common UD system. Remember, a polymer armor system composed of constituents that have
a higher wave speed (and identical toughness) will be able to move energy away from the impact site faster and will generally provide a better ballistic performance as compared to the slower wave speed system. This sort of exploratory project would be a great Masters thesis or nice short side project for a PhD student (of course the project could be vastly expanded to dig deeply into the failure mechanisms presented by the HB10 system, which could then push this study into a full-fledged PhD study). It would be interesting to see how the Cunniff equation performs in this new-age UD composite environment. As with most of Cunniff’s ideas, it would probably fare quite well\(^1\).

Finally, efforts must be made to finish out the work performed at APS regarding the strains developed in HB10 tape. Specifically, great care and effort should be made from a watchful eye in determining how much strain is developed in the the crystalline region of the material, thereby allowing one to back out how much strain is developed by the amorphous regions in the sample, all of which should be performed at an elevated strain rate. Although this may sound simple in nature (which it most likely is once the data has been gathered), this is in the author’s opinion, quite a difficult experiment, and requires the trained eye of a highly intelligent beam scientist. Very precise measurement must be made of both the macro strain measured from the Kolsky bar, and the crystalline strain (measured from the dynamic diffraction patterns). Care must be taken to ensure proper sample geometry, as nearly the entire experiment hinges on pure uniaxial tension, and an understanding of the sample geometry with respect to the beam and the detector. All in all, it would be a difficult, yet highly rewarding piece of information to gain for the study of polymers subjected to dynamic loading conditions.

As initially stated, please do ponder these thoughts with care. They have been written with some relevant background, but from an inexperienced researcher. Do

\(^1\)On this note, I am reminded of the first day in my advisor’s office. Professor Chen promptly told the student to read all of Cunniff’s work. I would highly recommend the same to any new researcher in composite armor systems
not let such recommendations sway any large scale action without first digging deep into the proper literature surrounding this ever expanding field.
REFERENCES
REFERENCES


[115] Iacp/dupont kevlar® survivors’ club.


Dyneema factsheet, 2008.


Spectra factsheet.


[152]


APPENDIX
A. Appendix

A.1 Dyneema® SK76 - Flat Projectile

A.2 Dyneema® SK76 - Round Projectile

A.3 Kevlar® KM2 - Flat Projectile

A.4 Kevlar® KM2 - Round Projectile
Figure A.2. 06221506a.
Figure A.4. 06221508a.
Figure A.5. 06231505a.
Figure A.6. 07291507d.
Figure A.12. 07301501d.
Figure A.13. 07301518d.
Figure A.14. 07311516c.
Figure A.15. 07311516c.
Figure A.16. 08011524b.
Figure A.17. 08011525b.
Figure A.18. 06231527a.
Figure A.19. 06231528a.
Figure A.21. 07281514a.
Figure A.22. 07281514b.
Figure A.23. 07281515a.
Figure A.24. 07281517a.
Figure A.26. 07281519a.
Figure A.27. 07281519b.
Figure A.28. 07291501d.
Figure A.29. 07301521c.
Figure A.30. 07311505c.
Figure A.32. 06231515a.
Figure A.33. 07281508Ka.
Figure A.34. 07281508Sb.
Figure A.35. 07281512Ka.
Figure A.37. 07301507d.
Figure A.38. 07311506c.
Figure A.41. 08011519b.
VITA
VITA

Matt Hudspeth is currently a doctoral graduate student in the Impact Science Lab at the School of Aeronautics and Astronautics of Purdue University. He also attended Purdue for both a Master’s degree in Materials Science Engineering and a bachelor’s degree in Mechanical Engineering. His PhD thesis research focus is comprised of analyzing the effectiveness of bullet proof vests via mechanical testing and impact on hierarchical size levels, namely fiber, yarn, and woven fabrics. He is also keenly interested in developing new experimental techniques which allow for in-situ analysis of material deformation and failure during high-rate loading with the most relevant advancement being made in phase contrast imaging and diffraction of materials being loaded in a Kolsky bar apparatus while placed in a synchrotron radiation source.

During his graduate studies, Matt was a NDSEG fellow, allowing for high levels of collaboration on simultaneous projects, ultimately resulting in the authorship of 17 publications from programs funded by various governmental agencies such as Army PEO Soldier, AFOSR, ONR and the DoJ. Regarding his main research topic, Matt’s success in body armor research was honored with the Gary Cloud Scholarship from the Society of Experimental Mechanics. Additionally, he led a team in winning of the National Institute of Justice Body Armor Challenge, resulting in a multi-student research program funded by the DoJ. As a main side project, Matt led a team to establish an integrated technique coupling high-rate imaging/diffraction of material subjected to high strain-rate loading, which has ultimately led to the award of routine beam time at the Advanced Photon Source of Argonne National Lab. Due to academic success, Matt was the recipient of several awards including the Jelinek, McKinley, Hine, and Merit Scholarships from Purdue University. He also stays active in professional societies, regularly aiding as a conference chair and was the winner of the the 2012 SEM student paper competition. Throughout graduate
studies, Matt has placed a high level of importance on the growth of undergraduate student researchers, regularly taking on summer students and leading them through the basics of hands-on experimental testing and design, resulting in seven students pursuing graduate school. Additionally, his leadership of undergraduate researchers was awarded the 2014 SURF Graduate Mentor of the Summer Award. Ultimately, from the research and leadership experiences gained throughout his graduate studies, it has become Matt’s goal to lead a world class research team that concentrates on providing solutions to national security problems with a stalwart emphasis focused on understanding fundamental mechanisms governing the physics of applied problems.

For his immediate future endeavors, Matt has accepted an offer for the President Harry S. Truman Fellowship in National Security Science and Engineering at Sandia National Laboratories in Albuquerque, New Mexico. Such an opportunity promises to provide a three year postdoctoral fellowship tenure allowing for high risk research unbridled by the typical need to gather financial stability, as the position is funded by internal lab support. Although daunting, he hopes he can live up to the Truman charge of providing ”an exceptional service in the national interest”.

A.5 Education

- PhD School of Aeronautics and Astronautics, Expected: May 2016 Purdue University, West Lafayette Cumulative GPA 4.0/4.0
- M.S. School of Materials Engineering, May 2012 Purdue University, West Lafayette Cumulative GPA 4.0/4.0
- B.S. School of Mechanical Engineering, May 2010 Purdue University, West Lafayette Honors Engineering Program Cumulative GPA 3.89/4.0 Graduated with Distinction

A.6 Fellowship

A.7 Journal Publications


16. **Matthew Hudspeth**, Jou Mei Chu, Boonhim Lim, Emily Jewell, Ernest Ytuarte, Weinong Chen, James Zheng. Effect of projectile nose geometry on the

A.8 Journal Publications (in review)


A.9 Journal Publications (in preparation)


4. Matthew Hudspeth, Zherui Guo, Niranjan Parab, Ted Danielson, Harm van der Werff, Weinong Chen. Effect of strain rate and orientation on Dyneema


### A.10 Work Experience

- Impact Science Lab, Fall 2010 - Present
  - Mechanical characterization and failure analysis of high-performance yarn and fabric via transverse impact with a high-velocity powder gun and light gas gun systems
  - Characterization and testing of high performance ballistic fibers (i.e. transverse compression, low/high rate tension, shear modulus testing, biaxial failure criterion)
  - Material mechanical characterization via Kolsky bar high-rate loading, with an emphasis on low force output materials (e.g. high-performance fibers)
  - Analysis of material high-rate deformation and failure using Phase Contrast Imaging and X-ray diffraction via synchrotron radiation from the Advanced Photon Source at Argonne National Lab

- India Institute of Science (IISC) Visiting Scholar, Summer 2010 - *Invited by Prof. Chandrasekar (Purdue University and IISC)*

  **Deformation field in plane strain flat punch indentation**
  - Nanoindentation and electron microscopy of surface deformation due to flat punch indentation
• Center for Materials Processing and Tribology, Fall 2008-2010

Controlled gradation of near surface layers in severe plastic deformation

– Electron microscopy of near surface deformed layers in ultrafine-grained copper and correspondence to nanoindentation measurements

• National Science Foundation Summer Undergraduate Research Fellowship (NSF-SURF) at Purdue University, Summer 2008

Thermal stability and ductility in high strength nanostructured alloys

– Microstructure characterization of nanostructured aluminum alloy particulate using transmission electron microscopy
– Strength measurements and microstructure observations during second-phase precipitation (nucleation, growth) in fine-grained aluminum alloys
– Sample prep using focused ion beam processing / electrolytic thinning

A.11 Peer-Reviewed Competition Awards

– Matthew Hudspeth, Ben Claus, Niranjan Parab, Weinong Chen
  Energy Signature Test (VEST)
  Submitted to National Institute of Justice Body Armor Challenge
  Result: 1st place winners
– Matthew Hudspeth, Xu Nie, Weinong Chen, Randolph Lewis
  Mechanical Properties of Major and Minor Ampullate Silks from the Spider Nephila Clavipes.
  Submitted to SEM International Student Paper Competition
  Result: 1st place winner

A.12 Achievements

NDSEG Fellow
Frank T. Jelinek Scholar
Thomas L. McKinley Scholar
Thomas Hine Scholar
Purdue Merit Scholar
Hoosier Scholar
SEM Cloud Scholar
SEM Student Paper Comp Winner
Purdue SURF Graduate Mentor Award