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Fatigue performance improvement in AISI 4140 steel by dynamic strain aging and dynamic precipitation during warm laser shock peening

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Received 30 July 2010; received in revised form 13 October 2010; accepted 13 October 2010
Available online 8 November 2010

Abstract

Warm laser shock peening (WLSP) is a thermomechanical treatment technique combining the advantages of laser shock peening and dynamic strain aging (DSA). Through DSA, WLSP of steel increases the dislocation density and stabilizes the dislocation structure by pinning of mobile dislocations by carbon atoms. In addition, WLSP generates nanoscale carbide precipitates through strain-induced precipitation. The carbide precipitates stabilize the microstructure by dislocation pinning. This results in higher stability of the dislocation structure and thus improves the stability of the compressive residual stress. In this study the mechanism of fatigue performance improvement in AISI 4140 steel by WLSP is investigated. It is found that microstructures formed after WLSP lead to a higher stability of dislocation structures and residual stress, which are beneficial for fatigue performance.

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Keywords: Warm laser shock peening; AISI 4140 steel; Dynamic strain aging; Dynamic precipitation; Carbide

1. Introduction

As a superior surface processing technique, laser shock peening (LSP) has been successfully used to improve the fatigue performance of metallic components [1]. By generating a work hardened layer and introducing compressive residual stress in the material surface the speed of crack initiation and propagation during cyclic loading is slowed down, which results in a fatigue performance improvement. LSP is an effective way to improve surface hardness, fatigue performance, corrosion resistance and wear resistance [2].

Steels are widely used in industry. LSP of steel has been extensively studied in the literature. For example, Nikitin [3,4] compared the near surface microstructure change and fatigue life improvement of AISI 304 stainless steel after LSP and deep rolling (DR). Hu [5] investigated LSP of AISI 1045 steel by ANSYS, validated by experiment. Chu [6] compared the microstructure, hardness and residual stress generated by LSP, DR and shot peening (SP) on Hadfield manganese steel. In Chu’s study it was found that LSP resulted in a large hardness increase due to the formation of a high density ε-martensite phase.

However, the compressive residual stress generated by surface processing techniques (SP, LSP, DP, etc.) is not stable during cyclic loading [7,8], especially at high testing temperatures [3,4,9,10]. For example, Altenberger et al. [11] investigated the thermal stability of the compressive residual stress and surface nanostructure generated in AISI 304 stainless steel and Ti64 alloy by dynamic precipitation (DP) and LSP by in situ transmission electron microscopy (TEM) study. It was observed that complete residual stress relaxation at 550–600 °C was related to the thermal instability of the near surface microstructure. In this way, the effect of fatigue life improvement by LSP is limited. Thus, it is very important to stabilize the microstructure and the compressive residual stress generated by LSP.

Dynamic strain aging (DSA) and DP can both improve the microstructure stability of metallic materials. DSA [12,13], the diffusion of C (carbon) and N (nitrogen) atoms
to dislocation cores in the temperature range 150–300 °C, is an important strengthening mechanism [14] in steel. In DSA the interaction between dislocations and solute atoms results in repeated pinning of dislocations and thus leads to enhanced work hardening [13,15]. At the DSA temperature the solute atoms (carbon and nitrogen) migrate to dislocation cores, which form so-called Cottrell clouds [16] in steel. The Cottrell clouds exert a pinning force on dislocations and inhibit dislocation movement during plastic deformation. For plastic deformation to continue, new mobile dislocations must be generated. This leads to dislocation multiplication and results in a higher dislocation density and a more uniform dislocation arrangement. Substantial efforts have been made to take advantage of DSA in treating steel. For example, Chen [17] improved the fatigue performance of AISI 304 stainless steel by plastic deformation at the DSA temperature. Kerscher et al. [18] increased the fatigue limit of SAE 52100 steel by TMT at the DSA temperature, and identified the optimal temperature (335 °C) that led to best fatigue performance improvement. Huang et al. [19] compared the fatigue performance of SA533B3 steel at room temperature and 300 °C and found that the better fatigue performance at 300 °C was a combined effect of DSA and the formation of carbide precipitates during cyclic loading.

Dynamic precipitation during hot deformation is also known as strain-induced precipitation (SIP). Dynamic precipitation differs from static precipitation in that the former results in the formation of nanoscale precipitates dynamically during warm deformation. In dynamic precipitation the dislocations generated by deformation act as favorable nucleation sites to grow precipitates dynamically. Compared with static precipitation, dynamic precipitation is more efficient in strengthening in that it takes a much shorter time to reach peak hardness. Tiitto et al. [20] investigated the effect of dynamic precipitation in steel on the hot flow behavior of alloy steel. It was found that the peak pinning force resulting from dynamic precipitation leads to a peak in the flow curve during hot deformation. As discussed earlier, DSA can increase the dislocation density generated by deformation. The high density dislocations, in turn, can provide numerous potential nucleation sites for dynamic precipitation. Thus, the effectiveness of DP can be improved through DSA. Liao et al. [21] proposed a nucleation mechanism to explain the ultrahigh dense nano-precipitation during WLSP, and found that dislocations after high strain rate deformation and elevated temperatures are the two most important factors. The nucleation model was validated by experiments.

The performance of surface processing techniques, including LSP, DR and SP, can be improved by taking advantage of DSA and DP. Matlock [15] compared the effect of DR of AISI 4140 steel at room temperature and 260 °C (DSA temperature). It was found that DR at the DSA temperature significantly increased the core hardness and also led to a more stable dislocation structure and thus improved the fatigue performance. High temperature DR of aluminum alloys was also proven to be more effective in fatigue performance improvement than room temperature DR by Juierm [22–24]. Harada [25] compared shot peening of spring steel at room temperature and elevated temperatures (100 °C, 200 °C, 300 °C and 400 °C). It was found that SP at the optimal treatment temperature (200 °C) tends to increase the near surface compressive residual stress magnitude and hardness due to the decrease in flow stress at high temperature. In addition, it was found that the magnitude of the residual stress generated by SP decreased due to recovery at treatment temperatures higher than 200 °C. Though it was not mentioned by Harada, the increase in hardness at 200 °C (in the DSA temperature regime) could also be partially attributable to DSA, which led to the pinning of dislocations by Cottrell clouds and resulted in a higher dislocation density and greater work hardening. In the warm shot peening work on AISI 4140 steel carried out by Wick [26] and Menig and Schulze [27] it was demonstrated that SP at elevated temperature (around 300 °C) improved the residual stress stability and led to better fatigue performance. According to Wick [26], in the warm peening samples static and dynamic strain aging occur simultaneously during and after warm peening, which leads to a higher surface hardness. In addition, DSA in warm shot peening leads to the formation of a high density of dislocations and more uniform dislocation arrangement, which contribute to a higher residual stress stability during cyclic loading.

As a superior surface processing technique LSP can also take advantage of TMT by treating steel in the DSA temperature regime (150–300 °C). Thus, it is of interest to investigate the effect of treating temperature on the fatigue performance improvement by LSP. In a previous study by our group [28] it was found that warm laser shock peening (WLSP) can significantly improve the stability of the compressive residual stress in AA 6061 alloys through the pinning of dislocations by the formation of a high density of nanoscale precipitates generated by dynamic precipitation. In this work WLSP of AISI 4140 steel was carried out and its effects on fatigue performance were studied. The microstructure of the samples treated after LSP and WLSP was characterized by transmission electron microscopy (TEM). The residual stress and dislocation density were measured by X-ray diffraction.

2. Experiments

2.1. Materials

Samples were cut and machined from a AISI 4140 steel plate with the chemical composition 0.41 C, 0.21 Si, 0.83 Mn, 0.025 P, 0.027 S, 0.91 Cr, 0.18 Mo, the remainder Fe (all wt.%). The sample dimensions were 76.2 × 10 × 2.38 mm. Before LSP the samples were austenitized for 20 min at 850 °C, oil quenched down to 25 °C, tempered at 450 °C for 2 h and cooled in a vacuum furnace. This procedure results in steel with a Vickers hardness of
310 VH and a microstructure of tempered martensite (Fig. 4).

2.2. Warm laser shock peening experiments

A schematic of the WLSP process is shown in Fig. 1. BK7 glass was used as the confining medium due to its high shock impedance and high melting point, making it suitable for LSP at elevated temperatures. In this case water cannot be used as the confining medium due to its low evaporation point. In practice, silicone oil (type 710) could also be used for confinement, due to its high vapor point (~300 °C) compared with water. Thin aluminum foil is used as an ablative coating material to protect the target material from surface melting. The working temperatures for WLSP are manipulated using a hot plate. A thermometer is used to monitor the sample temperature. The laser beam size used is 1 mm. The overlap ratio is 75%. Further details of the WLSP experiment can be found in Ye et al. [28].

2.3. Characterization

2.3.1. Micro-hardness

The micro-hardness change of the samples before and after LSP or WLSP is measured using a Leco M-400-H micro-hardness test machine with a 200 g load and a 10 s holding time. The average of five measurements was used for each data point.

2.3.2. Residual stress

A Bruker D8-Discover X-ray micro-diffraction system was used to measure the residual stress of the sample. The X-ray collimator used in this work is 0.1 mm in diameter. The [2 2 0] peak was used for stress analysis, which corresponds to a 20 angle of 123.916° in the unstressed state. The interference lines of the steel phase were determined at 11ψ angles from −50° to +50° using Co Kα1 radiation and analyzed by the sin²ψ method [29]. The X-ray peak broadenings were evaluated from the full width at half maximum (FWHM) integral values after removal of the Kα2 signal. The FWHM value at the 90° X-ray incidence angle of the Bragg diffraction [2 2 0] peaks was used as a measure of the relative dislocation density [29], or work hardening rate.

To measure the core residual stress the material was removed layer by layer by an electrolytic polisher (Proto Manufacturing Inc.). The electrolytic polishing medium was the AI solution from Proto Manufacturing Inc. To investigate the thermal stability of the compressive residual stress the samples were put in a furnace at 350 °C for different annealing times and then the residual stress measured. To investigate the cyclic stability of the compressive residual stress the residual stress was measured after different numbers of rounds of cyclic loading.

2.3.3. Tem

The TEM samples were prepared by the focused ion beam (FIB) lift-out method [30] in a FEI NovaLab 200 FIB system. TEM was carried out in an FEI Titan operated at 300 keV.

2.3.4. Fatigue test

A 100 KN MTS servo-hydraulic fatigue testing machine was used to carry out the three-point bending fatigue test, in load control mode. The loading profile is a sine wave function with a frequency of 5 Hz. The stress ratio R is 0.1 for all the fatigue tests (i.e. \( R = \sigma_{\text{min}} / \sigma_{\text{max}} \), where \( \sigma_{\text{min}} \) is the minimum stress and \( \sigma_{\text{max}} \) is the maximum stress). The maximal bending stress was calculated by \( \sigma = \frac{3PL}{2bh^2} \), where \( P \) is the applied load, \( L \) is the span for the bending fatigue test set-up, \( b \) is the width of the specimen and \( h \) is the thickness of the specimen. All the tests were carried out at room temperature and in a laboratory environment.

3. Results and discussion

3.1. Process conditions for warm laser shock peening

3.1.1. Laser processing condition

One of the most important parameters in LSP is laser intensity, which controls the shock pressure. In this study BK7 glass (shock impedance 1.44e6 g cm⁻² s⁻¹ [31]) was used as the confining medium, which has a much higher shock impedance compared with water (shock impedance 0.16556 g cm⁻² s⁻¹ [32]). According to Fabbro et al. [33] the laser-induced shock pressure could be estimated as:

\[
P(\text{GPa}) = 0.01 \sqrt{\frac{G}{2\pi r}} \sqrt{Z(\text{g/cm}^2 \text{s})} \sqrt{I_0(\text{GW/cm}^2)},
\]

where \( Z \) is that portion of absorbed energy contributing to the thermal energy of the plasma and \( Z = \frac{1}{\rho \sigma D} \) is the reduced shock impedance between the target material (steel 4140 shock impedance 3.96 g cm⁻² s⁻¹) [34], estimated as \( Z = \rho D \), where \( \rho \) is the material density and \( D \) is the shock velocity. From our calculations the shock pressure using BK7 as the confinement was about 2.7 times higher than that using water as the confinement.

In this study the laser intensities used were from 1.5 to 4 GW cm⁻² with a 0.5 GW cm⁻² interval. It was found that the confining medium (BK7 glass) cracked at laser intensities above 4.0 GW cm⁻². The residual stresses for
Laser intensities from 1.5 to 4.0 GW cm\(^{-2}\) under LSP and WLSP conditions were measured (Fig. 2). The estimated peak plasma pressure at different laser intensities were also plotted based on Fabbro’s model [33] (see Fig. 2). It was found that the residual stress magnitudes increased almost linearly with increasing laser intensity for both LSP and WLSP from 1.5 to 4.0 GW cm\(^{-2}\). In addition, the residual stress magnitudes for LSP and WLSP (250 °C) are very close at all laser intensities. The compressive residual stress magnitudes reach around 500 MPa for both LSP (501 MPa) and WLSP (519 MPa) at 4 GW cm\(^{-2}\). While a high magnitude of compressive residual stress is beneficial for fatigue performance, 4 GW cm\(^{-2}\) was chosen as the laser intensity in the following experiments in this study.

According to the study by Juijerm [23], the magnitude of the residual stress generated by deep rolling at high temperature (250 °C) is much lower than that at room temperature (50 compared with 260 MPa). Thus it is worth mentioning that the magnitudes of compressive residual stress are very close between LSP and WLSP, i.e. WLSP did not reduce the magnitude of residual stress compared with LSP. However, what is more important is the stability of residual stress, which will be addressed later.

3.1.2. WLSP working temperature

It is necessary to determine the optimal working temperature for WLSP in terms of compressive residual stress magnitude and hardness improvement. According to warm SP work on AISI 4140 steel by Menig and Schulze [27] an optimal peening temperature of 300 °C was identified. Considering that the DSA temperature of medium carbon steel is between 150 °C and 300 °C, temperatures from 100 °C to 350 °C with an interval of 50 °C were tested in this study. It was found that LSP at all temperatures leads to an improvement in hardness compared with LSP at room temperature (see Fig. 3). For all experiments below 300 °C the hardness increases with increasing temperature. This is because higher temperatures lead to a higher mobility of the solute atoms and thus more efficient DSA [17].

Fig. 2. Surface residual stresses for LSP and WLSP (250 °C) at different laser intensities and corresponding peak plasma pressures.

Fig. 3. Hardness at different temperatures (laser intensity 4 GW cm\(^{-2}\)).

Fig. 4. Initial microstructure of quenched and tempered steel 4140 without peening showing (a) retained martensitic laths and (b) Fe\(_3\)C cementite precipitates.
hardnesses at 250 °C (416 VH) and 300 °C (418 VH) are very close to each other. At 350 °C there is a drop in hardness, which could be caused by two effects: (1) 350 °C is higher than the upper range of the DSA temperature for steel 4140 and (2) thermal relaxation by dynamic recovery leads to a hardness drop. To obtain the greatest strengthening effect and to avoid dynamic recovery, 250 °C is used as the working temperature for WLSP. The magnitudes of compressive residual stress after LSP (room temperature) and WLSP (250 °C) are very close (Fig. 2) at different laser intensities (1.5–4.0 GW cm$^{-2}$).

3.2. Microstructures induced by DSA and DP

Microstructures are important in that they greatly affect the material properties. The initial microstructure of quenched and tempered steel contains retained lath-type martensites (Fig. 4a) and low density of lath-type precipitates (Fe$_3$C type) (Fig. 4b). DSA and DP are known to change the material microstructure, improving the properties. The microstructures of the LSP and WLSP samples are analyzed and compared here to determine the effect of DSA and DP.

3.2.1. Effects of DSA

The TEM micrographs (Fig. 5) show the microstructures on the LSP samples at different magnifications. In the LSP sample pile-up of localized dislocations and lamellar dislocation boundaries (indicated by the arrows in Fig. 5b and c) [35,36] can be observed. A detailed view (high magnification) of these dislocation bands is shown in Fig. 5d. These dislocation pile-ups are also called shear bands, which form when metallic materials are subjected to high strain rate deformation [37]. For the WLSP sample there are fewer dislocation pile-ups and more tangled dislocations [15] are observed (Fig. 6), which are more stable than the dislocation structure in the LSP sample. This is caused by DSA, which leads to enhanced dislocation multiplication and thus the formation of more tangled dislocations. Through DSA, the mobile dislocations are pinned by the solute atoms. The generation of new mobile dislocations is necessary for plastic deformation to continue. In WLSP of steel 4140 the solute atoms, especially carbon atoms, can diffuse quickly enough to form a Cottrell atmosphere at elevated temperatures during plastic deformation. For the pinned dislocation to break away from the Cottrell atmosphere greater stress is needed, which will result in the

![Fig. 5. TEM images of a LSP sample (4 GW cm$^{-2}$ at room temperature) showing lamellar dislocation bands at different magnifications.](image-url)
activation of other dislocation sources and the generation of new mobile dislocations. In this way, dislocation multiplication is enhanced, thereby leading to a higher dislocation density and more dislocation tangles [15,17].

The relative dislocation density depth profiles (expressed as the FWHM value of the \{2 2 0\} X-ray diffraction peak) for LSP and WLSP are compared in Fig. 7. It shows that WLSP leads to a higher dislocation density compared with LSP. This is another indication of the high dislocation density in the WLSP sample caused by DSA.

3.2.2. Effect of heat-assisted DP

During plastic deformation of steel dislocation cores with high carbon atom concentrations are generated, which act as potential nucleation sites for carbide precipitates to grow. These carbides hinder dislocation movement during cyclic loading by pinning the dislocations, which will improve the stability of the dislocations and also the stability of compressive residual stress. The microstructures of the LSP sample are shown in Fig. 8. In Fig. 8a and b a number of precipitates are observed near the subgrain boundaries, while few precipitates are observed within the subgrains. This is due to the high carbon atom concentration near the subgrain boundaries. Fig. 8c shows a TEM image of a precipitate at higher magnification.

Compared with those in LSP, precipitates were nucleated at a higher density in WLSP because the elevated tem-
perature during the high strain rate deformation increased the nucleation rate. The precipitation kinetics are strongly accelerated by concurrent deformation at high temperature in WLSP. In addition, DSA results in a higher dislocation density, resulting in more nucleation sites for nanoscale carbide precipitates. Fig. 9 shows TEM images of precipitates and dislocations after WLSP. During high strain rate deformation in WLSP high density dislocations are generated through DSA. At high treatment temperatures (250 °C) the speed of movement of carbon atoms is faster than at room temperature. Through diffusion, numerous carbon atoms migrate into the dislocation cores, which results in high carbon atom concentration in these regions. Thus, the dislocation cores formed by deformation become potential nucleation sites for carbide precipitates to grow. Due to the high treatment temperature in WLSP, more nucleation sites are generated compared with LSP. This results in highly tangled dislocations and precipitates, as shown in the dark field TEM image (Fig. 9a). Due to the high dislocation density, some of the nanoscale precipitates cannot be clearly observed in Fig. 9a. However, the diffraction pattern in Fig. 9b clearly indicates the presence of precipitates. The major spots in the diffraction pattern indicate a bcc (body-centered cubic) crystalline structure obtained along the 1 1 0 zone axis of the matrix material. Note that two neighboring grains (Fig. 9a, one big grain in the center, the other at the bottom) with slight misorientations contribute to the major spots in the diffraction pattern (Fig. 9b). The dimmer spots in between the major spots (circled by the rings) are attributed to the carbide precipitates (M23C6 type). As seen from the TEM image of the non-peened sample (Fig. 4b), lath-like precipitates (Fe3C) exist in the material after quenching and tempering. These precipitates are also preserved in the WLSP sample (Fig. 9c). However, the carbide precipitates (M23C6) generated by WLSP are different in shape from the initial precipitates. As shown in the dark field image in Fig. 9d, the lath-shaped precipitates are almost unobservable. Instead, many globular precipitates formed during WLSP are easily observed. Except for a few large sized ones (around 25 nm), most of the globular precipitates are around 10 nm in diameter. Compared with the precipitates in the LSP
3.3. Nanostructures on the surface

The TEM images in Fig. 10a and b show the grain structure in the very top surface (around 10 μm from the top) of the LSP and WLSP samples, respectively. It can be seen that both the LSP (Fig. 10a) and the WLSP (Fig. 10b) samples show nanocrystalline surfaces. According to the Hall–Petch equation $\sigma_y = \sigma_0 + k d^{-1/2}$, where $\sigma_y$ is the material yield stress, $\sigma_0$ is the initial yield stress, $k$ is material constant and $d$ is the grain size, the material strength increases with decreasing grain size. Whereas the majority of cracks initiate from the surface, surface grain refinement and strength improvement will certainly lead to an improvement in fatigue performance.

3.4. Hardening effect of WLSP

The surface hardness of LSP and WLSP samples at different laser intensities is compared in Fig. 11. For all laser intensities investigated WLSP leads to a higher surface hardness than LSP. This is due to the formation of a higher dislocation density (Fig. 7) and carbide precipitates (Fig. 9) by WLSP, both of which lead to surface hardening. The core hardness for LSP and WLSP at a laser intensity 4 GW cm$^{-2}$ is compared in Fig. 12. It can be seen that the surface hardness increases from 310 VH (kg mm$^{-2}$) to 390 VH and 417 VH for LSP and WLSP, respectively. Interestingly, a hardness peak (443 VH) was observed at around 300 μm below the surface for the WLSP sample, while the LSP sample does not have such a peak.

This indicates a significant difference in the hardening behavior during LSP and WLSP. For LSP at room temperature only strain hardening contributes to the hardness increase. Since plastic strain generated by LSP decreases...
with depth, the strain hardening effect also decreases with depth. This is consistent with the dislocation density depth profile in Fig. 7. Thus, the hardness decreases with depth for LSP. For WLSP the hardening mechanisms include strain hardening, the effect of DSA and heat-assisted DP. For WLSP at the DSA temperature DSA leads to a higher dislocation density and the pinning of dislocations by solute atoms. The enhancement of dislocation multiplication leads to the formation of a large number of obstacles to dislocation movement [17], the existence of which will generate resistance to dislocation slip. In addition, heat-assisted DP during WLSP leads to the formation of carbide precipitates by strain-induced precipitation. The precipitates formed also lead to higher strength by dislocation pinning. The pinning of dislocations by dislocation obstacles and precipitates increases the resistance to dislocation slip during plastic deformation. Therefore, the significant hardness improvement after WLSP is mainly due to the combination of strain hardening, DSA and precipitate hardening.

It is very interesting to note that there is a hardness peak in the material subsurface (300 µm below the surface) in the WLSP sample. Material hardness increases through increases in dislocation density (strain hardening and DSA) and precipitation hardening (dynamic precipitation). For dislocation density it is clear that hardness increases with dislocation density. For precipitate hardening, however, both precipitate size and volume fraction affect the hardening effect. There is a critical size for precipitate particles to exert an optimal pinning effect on dislocation movement. Since the top surface has the highest dislocation density in the WLSP sample, the subsurface hardness peak must be caused by precipitate hardening. The dislocation density is highest at the material top surface, which results in the highest density of favorable nucleation sites for DP. In this way, high density nanoscale precipitates were generated. However, since the WLSP processing time is short (less than 10 min), the precipitates cannot grow large while they are competing for carbon atoms. For the subsurface of the WLSP sample the dislocation density is relatively lower, which might result in large precipitates, which are more effective in dislocation pinning. Further investigation is needed to confirm this. The complex interactions
between strain hardening, DSA and DP determine the material hardness. The processing conditions (laser intensity, temperature, post-deformation aging, etc.) and initial material conditions (heat treatment history) play a very important role in this aspect. However, optimization of these parameters is beyond the scope of this study, but is currently under investigation in another effort by our group.

3.5. Distribution of residual stress

The residual stress distribution depth profiles of LSP and WLSP are compared in Fig. 13. At the top surface, LSP and WLSP induce very similar residual stress magnitude values. In the subsurface within 100 μm WLSP produces a higher magnitude of compressive residual stress. For example, at 50 μm below the surface the magnitudes of compressive residual stress for LSP and WLSP are 315 and 465 MPa (47.6% higher than LSP), respectively. For regions below 150 μm the residual stress magnitudes are very close for LSP and WLSP. As discussed by Harada [25], higher processing temperatures tend to reduce the flow stress during deformation, which leads to an increase in plastic deformation. However, high treatment temperatures could also result in residual stress relaxation through dynamic recovery. The coupled effect of reduced flow stress and residual stress relaxation resulted in the residual stress depth profiles of LSP and WLSP shown in Fig. 13.

3.6. Thermal and cyclic stability of surface residual stresses

The stability of the compressive residual stress is at least as important as its magnitude. Many components and structures are used at elevated working temperatures, which requires that the compressive residual stresses generated by LSP have high stability at elevated temperatures. The residual stresses of the LSP and WLSP samples after annealing at 350 °C for different times are shown in Fig. 14a. It can be clearly seen that the WLSP samples have a higher thermal stability of residual stress. For example, after 500 min the compressive residual stress from LSP decreased from 524.8 to 330.6 MPa, which corresponds to a 37% decrease, while WLSP decreased by only 20.7%.

The cyclic stability of compressive residual stress is also important in that it determines the effectiveness of fatigue performance improvement. The residual stress versus number of cycles for LSP and WLSP are compared in Fig. 14b. From Fig. 14b it can be observed that with different numbers of cyclic loading the WLSP samples have a higher magnitude of compressive residual stress than LSP samples. For example, after 1 K cyclic loading the residual stress magnitude of the LSP samples decreased by 29.2%, while that of the WLSP samples decreased by 19.1%. This indicates a higher cyclic stability of the WLSP samples than the LSP samples. With a higher stability, the compressive residual stress generated by WLSP is more effective in decreasing the crack propagation speed than that generated by LSP.

Both the thermal stability and cyclic stability of residual stress are related to the material microstructures after LSP and WLSP. In the current case of thermal annealing at 350 °C no significant yield stress reduction occurs in steel.
4140, which rules out the possibility of residual stress relaxation through a reduction in yield stress. For thermal annealing at 350 °C both diffusional creep and dislocation thermal glide lead to stress relaxation [7]. For the WLSP samples peened at the DSA temperature a higher dislocation density and more diffuse dislocation tangles make diffusional creep less efficient compared with non-uniform dislocation pile-up (Fig. 5a) in the LSP samples. In addition, the dislocations are impeded by Cottrell clouds formed during WLSP. Also, the high carbide precipitate density resulting from DP during WLSP exerts a pinning force on the dislocations. The combined effect of pinning by Cottrell clouds and carbide precipitates increases the resistance to dislocation movement, which contributes to dislocation stability during thermal annealing and cyclic loading. The comprehensive effects of WLSP leads to a high thermal and cyclic stability of compressive residual stress compared with LSP.

3.7. Fatigue performance

Like other popular surface treatment techniques (SP and DR), the fatigue life improvement after both LSP and WLSP is more efficient for a high cycle regime (HCR) than for a low cycle regime (LCR). The stress–lifespan (S–N) curves for bending fatigue testing after various processing conditions are shown in Fig. 15. The S–N curve moves to the right after LSP and WLSP. Under certain stress magnitudes (1200 and 1500 MPa) the fatigue life after WLSP is 3–5 times higher than that after LSP. In addition, both LSP and WLSP can improve the bending fatigue strength of AISI 4140 steel, however, WLSP is more effective. In Fig. 15 the bending fatigue strength for an as-machined sample, LSP samples and WLSP samples are 875, 1125 and 1200 MPa, respectively.

LSP can improve the fatigue behavior by surface hardening (improving resistance to fatigue initiation) and the introduction of compressive residual stress (decreasing crack propagation speed). Besides the benefits of LSP, WLSP results in a greater hardness improvement, a higher dislocation density and a more uniform dislocation arrangement by DSA, and a higher precipitate density by DP, the combined effects of which result in higher residual stress stability. Compared with LSP, WLSP treatment increases the dislocation density and changes the dislocation structure to a more stable state by locking mobile dislocations by carbon atoms and carbide precipitates. The higher stability of the dislocation arrangement generated by WLSP leads to an increase in the stress amplitude needed to induce the movement of new mobile dislocations. Also, the resistance to crack initiation is increased by greater surface hardening and the crack growth rate decreased by more stable compressive residual stress. Thus, compared with LSP at room temperature, WLSP can greatly improve the fatigue performance of the samples.

4. Conclusion

In this paper the mechanisms of fatigue performance in AISI 4140 after LSP and WLSP were investigated. LSP improves the bending fatigue performance by surface work hardening and the introduction of compressive residual stress. In WLSP the effects of DSA lead to the formation of a uniform and high density dislocation arrangement. In addition, DP during WLSP leads to the formation of ultrafine carbide precipitates, which prohibit dislocation movement during cyclic loading. The combined effects of DSA and DP in WLSP stabilize the dislocation structure by locking mobile dislocations, which leads to better fatigue performance than after LSP. In addition, the thermal stability of the residual stress generated by WLSP is much higher than that generated by LSP, which indicates the beneficial effects of WLSP in high temperature fatigue performance.

Acknowledgements

G.J. Cheng appreciates the supports from the Office of Naval Research (ONR) through the Young Investigator Program and a NSF Grant (CMMI 0900327).

Reference