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The Metallography of Impact Fatigue

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THE METALLOGRAPHY OF IMPACT FATIGUE

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INTRODUCTION

Compressor valves are subject to two causes of failure due to the cyclic loads which they experience. The first type of failure is induced by the bending stresses and represents a case of conventional fatigue failure. As such, measures for the prevention of this type of failure are relatively well understood. The second type of failure involves the fracture of small chips from the edge of the valve and is termed impact fatigue failure as it is associated with the repeated impact of the valve against its seat.

The phenomenon of impact fatigue has not been examined in detail and very little is known of the microstructural processes which lead to failure. Svenzon (1) has developed a special testing machine, the S.I.F.T., which allows impact fatigue failure to be initiated under controlled conditions and fatigue limits to be determined for various materials. The impact fatigue strength of several valve steels has been established using this machine (1). However, some experimental difficulty exists in determining the intensity of the impact when materials of different elastic and strength properties are tested.

Fractographic studies of the impact fatigue failure of valve steels (1) have shown that the fracture initiates below the surface of the specimen in the region between the valve edge and the contact area with the seat. The small effect of surface condition upon impact fatigue confirms the probability of subsurface crack nucleation. The first external signs of cracking were always perpendicular to the edge of the specimen. The region between these radial cracks then falls away to give the typical final appearance of an impact fatigue failure. Stress analysis (1) of the situation when a valve impacts a seat squarely shows that the stress levels are very low, of the order of 100 N/mm^2. However, larger shear stresses are produced if the valve impacts the seats obliquely. High-speed cine-photography has shown that, due to torsional and flexural vibrations, almost all impacts are oblique.

* *This work was performed while on leave at the Steel Research Centre Sandvik AB, Sandviken, Sweden.*

An attempt to examine the microstructural changes in valve steels upon exposure to impact fatigue (1) failed to reveal any observable changes. No change in hardness of the steels could be detected after impact fatigue.

The aim of the present investigation was to examine in greater detail the effects of impact fatigue upon the microstructure of the material and this paper reports the preliminary results obtained to date.

EXPERIMENTAL

Since the conventional valve steels are extremely complex in microstructure, a number of simpler materials were examined in an attempt to isolate the effects of impact fatigue. The materials examined were annealed commercial purity aluminium, annealed SANDVIK 5R60, Nitinol in the annealed austenitic condition and annealed SANDVIK 12R10. Samples of SANDVIK 20C valve steel which had failed by impact fatigue were also examined. Transmission electron microscope samples were prepared from the region between the contact ring and the specimen edge since this is the area of crack initiation. Comparative samples were prepared from the area of the sample which was clamped during testing. The samples were examined at 100kV in a JEM-100C electron microscope.

RESULTS

By adjustment of the intensity of impact, it was possible to produce impact fatigue failures in all of the materials examined. The difficulty in establishing an absolute measure of the intensity of the impact makes it impossible to derive any comparative values of the impact fatigue strength of the various materials.

Aluminium

Figure 1 shows typical dislocation arrangements in the aluminium samples, both for the region experiencing impact fatigue and the gripped and hence unstrained area. Figure 1(a) shows the structure of the material which has not experienced impact fatigue. The structure consists of well-developed subgrains and two large intermetallic particles can be seen, The subgrain boundaries are usually attached to these particles. The structure in the vast majority of the impacted samples was the same as in the unstrained regions. However, in

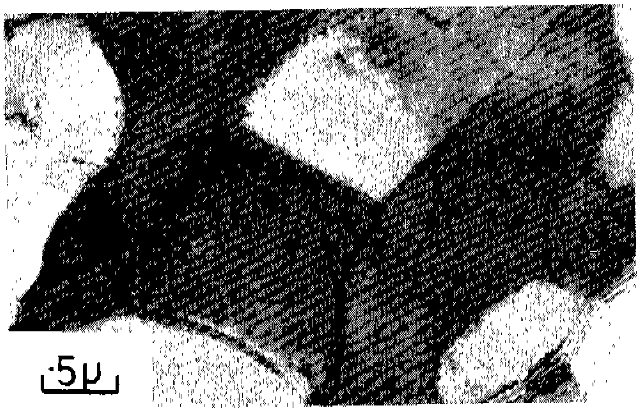

Figure 1(a)
An electron micrograph showing the subgrain structure of the unstrained region of the aluminium sample.

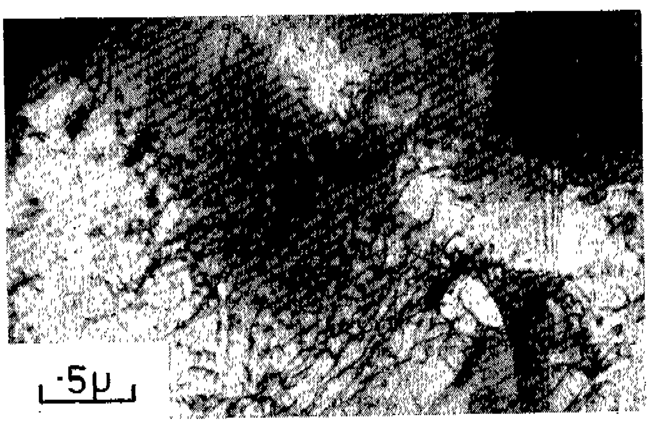

Figure 1(b)
An electron micrograph showing a region of increased dislocation density in the vicinity of a second phase particle in the impacted region of the aluminium sample.

some localized regions some effects of impact fatigue could be seen as shown in Figure 1(b). This shows a greatly increased density of tangled dislocations in the vicinity of a large second phase particle. A comparison of this figure with Figure 1(a), reveals a significantly increased dislocation density in the impacted material in localised regions. This was especially evident in the regions near intermetallic particles in the samples examined. Thus, the metallography of the aluminium impact fatigue samples indicates localized deformation in the region of some large intermetallic particles. The deformation is very inhomogeneously distributed and the great bulk of the material appears to have undergone little or no deformation.

SANDVIK 5R60

Figure 2(a) shows the microstructure of the unstrained regions of the samples. The structure is typical of a well annealed austenitic steel with a low dislocation density being present. Most of the areas of samples which had been subjected to impact fatigue had a similar structure to this with no signs of deformation. However, in some areas of some samples the very localized damage depicted in Figure 2(b) and (c) was found. No positive connection between these bands and any element of the microstructure could be established. The bands are related to the crystallography of the grains and follow (111) planes, and so change direction when crossing a grain boundary. Figure 2(c) shows a possible association of bands with a grain boundary particle. Within the intense deformation bands the dislocation structures are complex and some recovery appears to have occurred resulting in the formation of networks as shown in Figure 2(a).

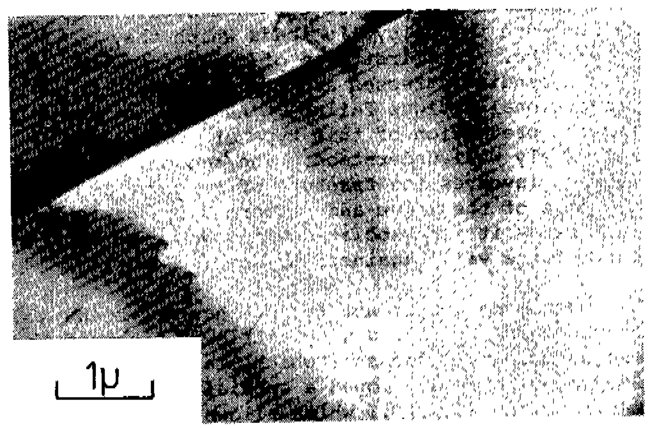

Figure 2(a)
An electron micrograph revealing the microstructure of the unstrained region of a SANDVIK 5R60 sample.

Figure 2(b)
An electron micrograph showing an intense deformation band in the impacted region of a SANDVIK 5R60 sample.

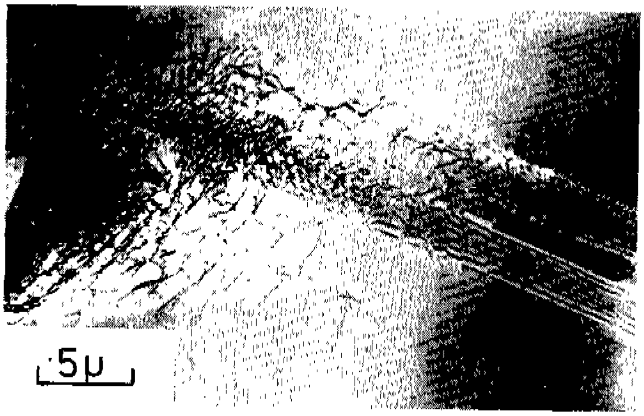

Figure 2(c)
An electron micrograph showing the possible association of a deformation band with a large grain boundary particle in SANDVIK 5R60.

The impact fatigue of 5R60 appears to result in very intense localized deformation. No positive evidence of the association of these intense bands of deformation with any element of the microstructure was obtained.

Nitinol

Nitinol was chosen for the investigation because of its extremely high damping capacity. Damping capacity had been identified in the investigation of a range of valve steels as an important property determining impact fatigue resistance (1). It was proposed that materials of high damping capacity suffered less oblique impacts due to the damping of the torsional and flexural vibrations and so suffered lower stress levels during impact. Unfortunately, the difficulty in establishing absolute values of the impact intensity precluded any assessment of the relative resistance of Nitinol to impact fatigue.

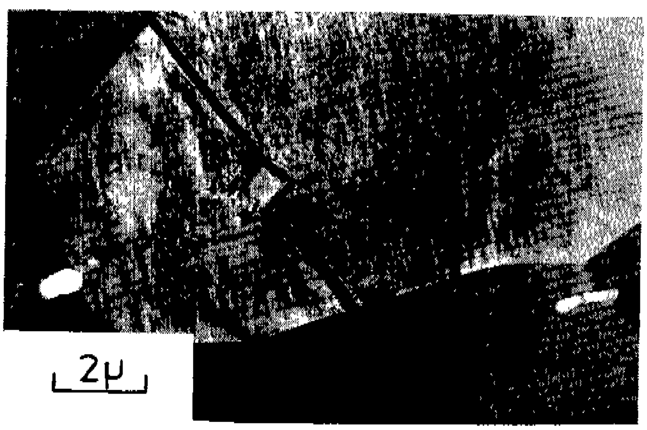

Figure 3(a)
A bright-field electron micrograph showing the association of a deformation band with a second-phase particle in impact-fatigued nitinol.

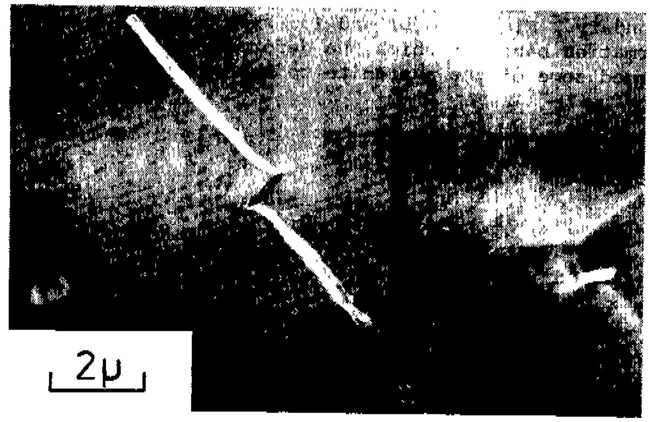

Figure 3(b)
The corresponding dark-field electron micrograph showing the presence of martensite in the deformation band in impact-fatigued nitinol.

Figure 3(c)
An electron migrograph showing dislocation arrays associated with large second-phase particles in impact-fatigued nitinol.

Figure 3(a) and (b) are a bright field-dark field pair of micrographs with the dark field image formed from a martensite reflection showing a localized deformation band which appears to have interacted with or originated at a second phase particle. Figure 3(c) shows another example of localized deformation in the form of dislocation arrays which are typical of austenite which has reverted from martensite (2). Again as in the other materials examined, the deformation was extremely localized and the majority of the sample appeared to be unaffected by the impact fatigue.

SANDVIK 12R10

In this material most areas of the foils examined showed no evidence of deformation and the impact fatigue damage was confined to narrow and isolated bands. Figure 4(a) is a low magnification micrograph of a deformation band in impacted 12R10. The crystallographic nature of the band can be seen in the directional changes when it crosses a grain boundary. Figures 4(b) and (c) show intense deformation bands in which the deformation has transformed some of the austenite to martensite.

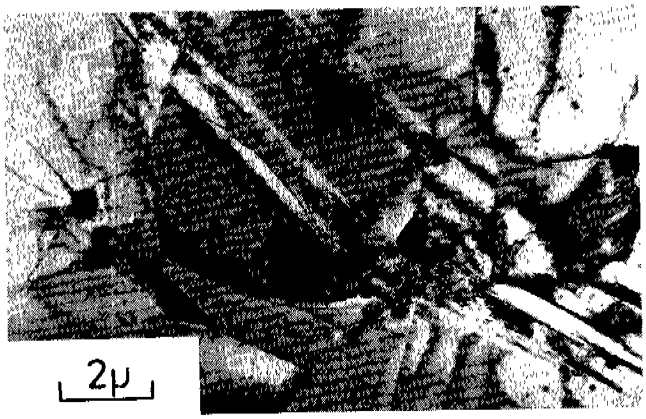

Figure 4(a)
An electron micrograph showing a deformation band crossing several grains in impacted SANDVIK 12R10.

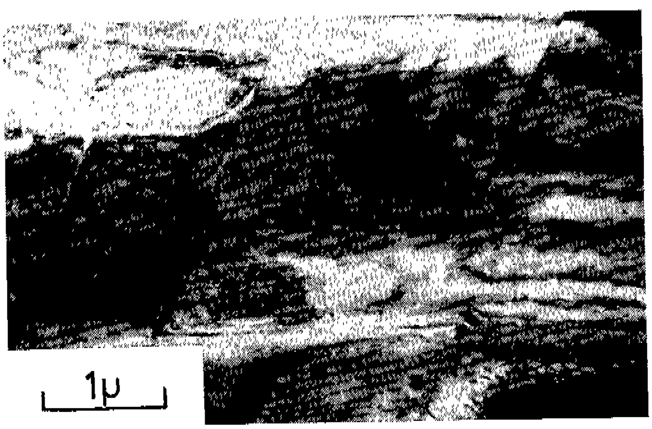

Figure 4(b)
An electron micrograph showing a deformation band in impact-fatigued SANDVIK 12R10.

No convincing evidence of a clear association of carbide particles with the initiation of deformation bands could be established as the origin of the bands was never in the viewable area of the foil.

SANDVIK 20C

Figures 5(a), (b) and (c) are micrographs of the structure of impact-fatigued 20C valve steel. The microstructure of this material consists of tempered martensite and relatively large carbide particles and is very complex. This complexity makes it difficult to be sure of the effects of impact fatigue. However, Figure 5(b) and (c) show a higher than average dislocation density in the vicinity of carbide particles. As such, the micrographs are not unlike those obtained for aluminium.

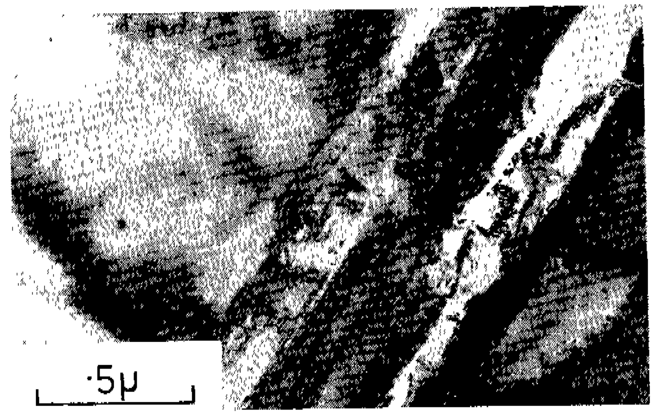

Figure 4(c)
An electron micrograph showing a deformation band in impact-fatigued SANDVIK 12R10.

Figure 5(a)
An electron micrograph showing the typical structure of SANDVIK 20C valve steel.

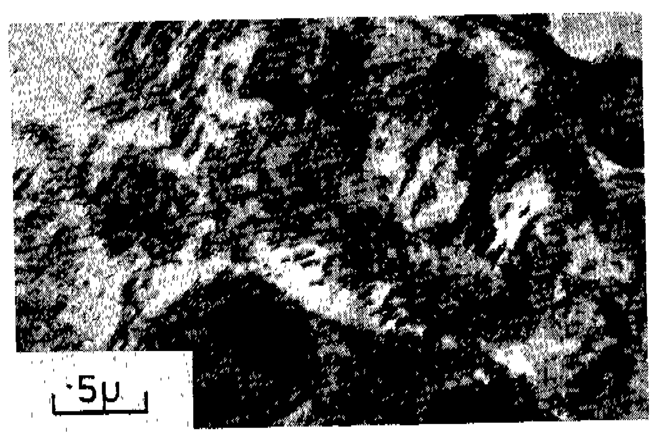

Figure 5(b)
An electron micrograph showing the microstructure of impact-fatigued SANDVIK 20C.

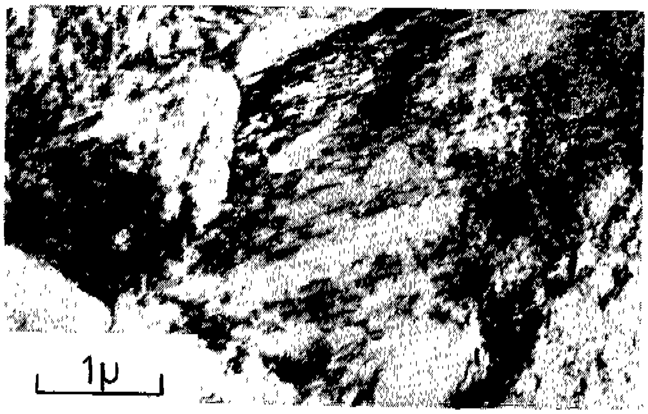

Figure 5(c)
An electron micrograph showing increased dislocation density in the vicinity of large carbide particles in impacted SANDVIK 20C.

DISCUSSION

The deformation which occurs during impact fatigue is of a highly localized nature. This probably explains why measurements of hardness before and after impact fatigue fail to show any significant change. The deformation is much more localized than that found in normal tension-compression fatigue where there is intense dislocation activity in persistent slip bands but also deformation in the surrounding matrix material (3). The very intense deformation could be likened to that observed near growing fatigue cracks in iron and stainless steels (4,5).

The presence of these localized regions of deformation probably indicates that the shear stress operating in these planes greatly exceeds that in the neighbouring regions. One can speculate that it is an interaction of the stress waves produced upon impact with some discontinuity in the material which locally causes high shear stresses to produce the band. There is some evidence for the association of the local areas of deformation with large particles so these particles may represent such a discontinuity. The lack of general deformation indicates that the average level of stress is low as indicated by the stress analysis performed by Svenzon (1). It seems likely that a stress analysis of a sample containing discontinuities may be necessary to establish the potential local stress levels involved in impact fatigue.

Thus, a possible explanation of the origins of impact fatigue failure involves the initiation of localised deformation at discontinuities in the matrix by the interaction of the stress waves with these discontinuities. Such an explanation appears to fit the experimental facts known about the phenomenon of impact fatigue failure.

CONCLUSIONS

1. Impact fatigue failure occurs in annealed commercial purity aluminium, annealed SANDVIK 5R60,Nitinol, and annealed SANDVIK 12R10.
2. Impact fatigue causes the formation of highly localised areas of deformation in these materials.
3. A possible explanation of the highly localised deformation involves the interaction of the stress waves produced by the impact with discontinuities in the microstructure of the material.

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