INFLUENCE OF SHOT PEENING ON THE FATIGUE BEHAVIOR OF TITANIUM ALLOYS

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ABSTRACT

The purpose of the work was to investigate the effect of shot peening on the change of the surface layer properties and their influence on the fatigue behavior of Titanium alloys in vacuum and aggressive environments. The influence of each of the important parameters (residual stresses, dislocation density and surface roughness) was evaluated separately. If the residual stresses remained stable during fatigue their influence on fatigue strength was characteristically different for push-pull compared to rotating beam testing. The results showed that residual stresses could increase or decrease the fatigue life. The stability of the residual stresses depended on the material cyclic hardening or softening behavior. Depending on many contributing factors crack nucleation occured at the surface or inside the material. If crack nucleation occured at the specimen surface the high dislocation density delayed nucleation whereas an increasing surface roughness had the opposite effect.

KEYWORDS

Titanium alloys; dislocation density; residual stresses; surface roughness; cyclic hardening and softening; cyclic stability of residual stresses; fatigue crack nucleation and propagation; environmental effects.

INTRODUCTION

The type of surface treatment of structural parts has a strong influence on the fatigue life because in fatigue the cracks usually nucleate in the surface layer. It is well known that especially shot peening leads to a remarkable increase in fatigue life. In general shot peening raises the dislocation density in the surface layer and because of the inhomogenious plastic deformation residual macrostresses are developed and further the surface topography is changed [1,2]. In addition it must be considered that the shot peening procedure is a fatigue process to the surface layer in which the peening pressure can be described qualitatively as the stress amplitude and the exposure time as the number of cycles in a fatigue test. It is often believed that the residual compressive stresses in the surface layer are the main reason for the improvement in fatigue life due to shot peening [3]. On an austenitic precipitation hardened steel it could be recently shown that the homogenization of the slip distribution in the surface layer due to a shot peening process has a marked influence on the fatigue crack nucleation [4].

EXPERIMENTAL PROCEDURE

The peening treatment was performed in a special apparatus that enables a defined shot peening treatment on round and flat specimens. The ball material was cast steel with an average ball size of 0.6 mm and a vickers hardness of 490. The number of balls per time could be kept constant for a wide range of peening pressures. The highest peening pressure of 12 bars resulted in ball velocities of nearly 40 m s^{-1} and an Almen intensity value of A = 0.35 mm.

The investigations were performed mainly on the Ti-6Aℓ-4V alloy and in addition on a binary Ti-8.6Aℓ alloy. For the Ti-6Aℓ-4V a coarse lamellar microstructure with a yield stress ($\sigma_{0.2}$) of 925 MPa, true fracture stress (σ_F) of 1050 MPa and true fracture strain (ε_F) of 0.15 was established in addition to a fine equiaxed microstructure with an average α-grain of 1-2 μm. In contrast to the lamellar microstructure having no crystallographic texture the equiaxed structure exhibited a strong texture of the hexagonal α-phase with basal and transversal portions (B/T) due to the unidirectional rolling process at 800°C [5]. From this textured material round smooth specimens were taken with the loading axis parallel (RD) and perpendicular (TD) to the rolling direction. The B/T-RD condition exhibited tensile properties of $\sigma_{0.2}$ = 1120 MPa, σ_F = 1650 MPa, ε_F = 0.62 and the B/T-TD condition of $\sigma_{0.2}$ = 1170 MPa, σ_F = 1515 MPa and ε_F = 0.55. After peening the dislocation density (ρ) was measured by broadening of X-ray interference lines of the $(21\bar{3}3)$ plane of the α-phase. The residual stresses (σ_R) and the stress distribution profile below the specimen surface after removal of surface layers by electropolishing were calculated by the $\sin^2\psi$ method [6]. The surface roughness (R_a) was measured by a profilometer. The change of the material strength in the surface layer due to shot peening was measured by microhardness. Fatigue tests of the electrolytically polished starting condition and the various shot peened conditions were done under push-pull loading in vacuum and 3.5 % NaCℓ solution and in addition with the rotating beam method (specimen diameter 3.6 mm) in laboratory air and dry argon atmosphere.

RESULTS AND DISCUSSION

Before fatigue testing the peening parameters were varied in a wide range (pressure from 2.5 to 10 bars, exposure time between 0.5 and 24 minutes) and the change in the surface layer properties was studied.

Due to the plastic deformation during shot peening the dislocation density ρ was increased. Whereas the maximum value at the surface seemed to be nearly independent of the peening pressure the plastic deformation zone (distance from the surface where ρ reached the original value) strongly increased with increasing peening pressure. For example with increasing pressure from 2.5 to 10 bars the plastic deformation zone was extended from 50 to 200 μm. Compared to peening pressure the effect of exposure time was not pronounced, that means for all peening pressures saturation of ρ is reached at short exposure times less than or equal to 4 minutes.

The distribution of the residual stresses for the coarse lamellar microstructure is shown in Fig. 1 (exposure time 4 minutes). The maximum compressive residual stresses and their depth below the surface increased with peening pressure.
For example peening with 2.5 and 10 bars resulted in a maximum compressive stress of 680 MPa at a depth of 25 μm and 750 MPa at a depth of 50 μm respectively. The measured depth of the compressive stress field was of the same order as the depth of the plastically deformed zone. Similar to the ρ-distribution the exposure time had little influence on the residual stress distribution. For the equiaxed structures similar profiles of the residual stresses were measured only with somewhat

higher (100 - 200 MPa) maximum values. Because the tensile residual stresses balancing the outer compressive stresses cannot be directly measured because the original stress equilibrium is changed due to the layer removal, tensile tests were done on shot peened specimens. Figure 2 shows clearly the decrease in yield stress compared to the unpeened condition due to the increasing interior tensile residual stress with increasing peening pressure. It is important to note that after an annealing treatment of 1 h at 500°C shot peened specimens exhibited the same stress-strain curve as the unpeened condition, that means the residual stresses were strongly relieved. X-ray measurements on peened and stress relieved specimens (1 h 500°C) showed nearly a complete stress relief from stress values at the surface of 800 MPa to about 200 MPa after the annealing treatment.

The surface roughness (R_a) starting with the electrolytically polished condition (R_a = 0.15 µm) strongly increased with peening pressure to a maximum value of R_a= 2.3 µm after peening with 10 bars. The roughness increased to a saturation value after short exposure times of 4 minutes.

Due to the fatigue process during shot peening the material strength will be increased with peening pressure and exposure times if the material cyclic hardens but decrease if cyclic softening occurs. Microhardness measurements of the surface zone showed a decrease in the material strength after peening with higher pressures or longer exposure times. Figure 3 shows an example by comparing the microhardness profile after peening times of 4 and 24 minutes. It can be seen that the microhardness decreased with increasing peening time. This softening behavior of the Ti-6Aℓ-4V alloy was also found in LCF tests [7]. The reduction in material strength occured although the dislocation density was nearly unaffected by longer exposure times and increased with peening pressure. For this evaluation it is important that the total material removal after peening for 24 minutes at 4 bars was only about 2.5 µm. To understand the fatigue behavior of shot peened specimens it is quite useful to separate the influence of each of the various parameters reported. Figure 4 shows the results of rotating beam tests in laboratory air of the condition B/T-RD. The influence of the residual compressive stresses on the fatigue life is given by comparison of the curves A and B. It should be pointed out that for B/T-RD specimens even in the peened conditions the fatigue cracks always nucleated at the surface. Due to the stress relief treatment in vacuum the fatigue strength was drastically reduced from 725 to 375 MPa. It should be noted that the dislocation density was nearly unchanged by the stress relief treatment. Comparing these results to the difference between the curves C and D (same conditions as A and B respectively but with an additional removal of a 20 µm thick layer from the surface by electropolishing) it can be seen clearly that in this case the effect of residual stresses was only pronounced if the specimen surface was rough and therefore easy crack nucleation took place at low stress amplitudes. The compressive stresses then had a marked influence on fatigue life because of the positive effect on crack propagation.

The influence of the surface roughness without the influence of the residual stresses is illustrated by the comparison of the curves D and B. The remarkable loss in fatigue strength of 425 MPa resulted from the easy crack nucleation at the rough surface.

The effect of a high dislocation density alone can be estimated by the difference between the curves A and B of Fig. 5. Due to the influence on crack nucleation the mechanical polishing ($Aℓ_2O_3$ paste with a particle size of about 10 µm) resulted in an increase in fatigue strength of 155 MPa over the electrolytically polished condition whereas the improvement at higher stress amplitudes where crack propagation played a dominant role became small. Because the plastic deformation zone was only of the order of 1 µm no line broadening of X-ray interference lines (depth of X-ray penetration was about 8 µm) could be found. Residual stresses in

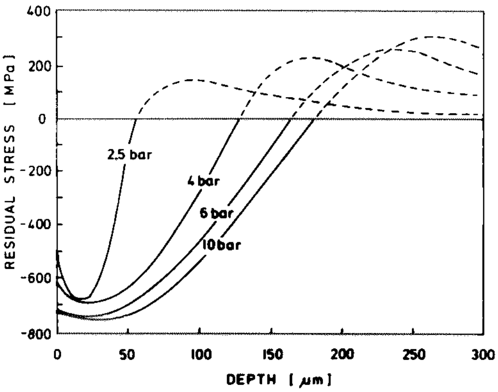

Fig. 1

Residual stress distribution after shot peening with different peening pressures (exposure time = 4 min.)

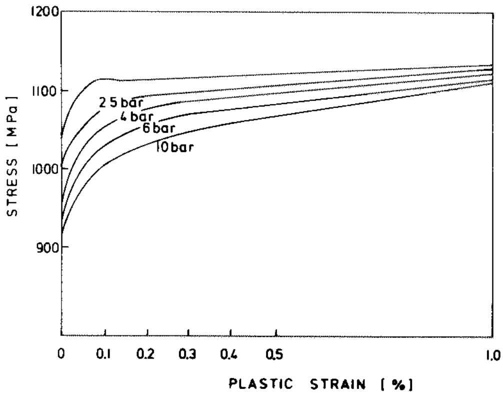

Fig. 2

Tensile tests, stress-strain curves for the unpeened and different peening conditions (exposure time = 4 min.)

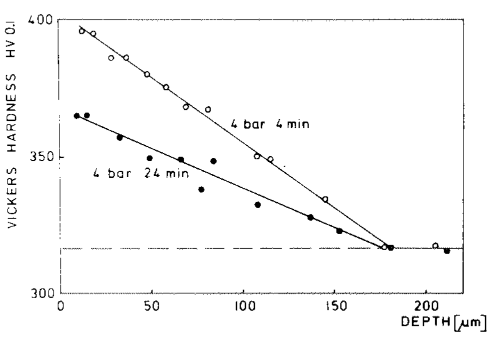

Fig. 3

Microhardness profile in the surface layer after shot peening for different peening times

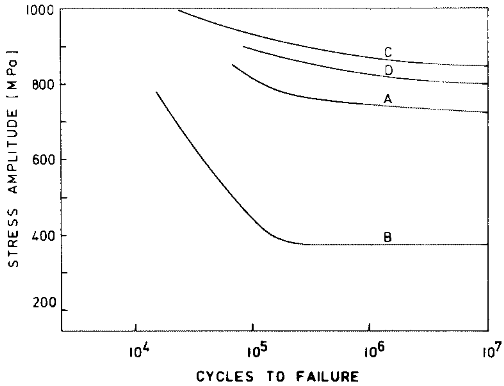

Fig. 4

Alloy condition B/T-RD, S-N curves for different surface conditions (rotating beam, air)

A: shot peened
B: A + annealed 1h 500°C in vacuum
C: A + 20 μm surface removal
D: B + 20 μm surface removal

the mechanically polished condition (smooth surface topography) were not pronounced because no decrease in fatigue strength was observed after a stress relief treatment of 1 h 500°C. It should be noted that the effect of mechanical polishing on the fatigue behavior seemed to depend on the grain size of the material. Polishing of the binary Ti-8.6Aℓ (grain size about 45 μm) resulted in a smaller increase in fatigue strength.

The influence of the fatigue softening during shot peening can be seen qualitatively by comparison curve A in Fig. 5 with curve D in Fig. 4. The effect of the high dislocation density induced by the shot peening process can be evaluated by comparison curve D of Figure 4 with curve B in Figure 5.

How much the residual compressive stresses which are developed during shot peening can influence the fatigue behavior essentially depends on their cyclic stability [1,8]. It can be assumed that the cyclic stability of the residual stresses is depending on the fatigue softening or hardening behavior. Figure 6 shows the difference in the cyclic stability of the residual surface stresses between shot peened specimens of B/T-RD (curve B) and B/T-TD (curve A) tested in compressive bending. The higher cyclic stability of the residual stresses in B/T-TD was in a good agreement with the much lower softening behavior compared to B/T-RD observed in LCF tests [7]. In the following the fatigue results of shot peened specimens of these different conditions are discussed with regard to the cyclic stability of the residual stresses.

The fatigue life of B/T-TD at a constant stress amplitude for push-pull tests under vacuum conditions as a function of peening pressure are given in Fig. 7. Since the residual stresses were less reduced during fatigue the increasing tensile peak stress with increasing peening pressure (see also Fig. 1) led to an earlier crack nucleation and therefore to a decline in fatigue life. The observed depth of the crack nucleation sites (Fig. 8) could be well correlated to the depth of the residual tensile peak stress. In a push-pull test the crack propagation was therefore nearly independent of the compressive stresses in the surface layer, because the cracks could unhindered propagate to the interior of the specimen due to the homogenious applied stress.

Compared to the push-pull results (curve C in Fig. 9) the shot peening treatment led to a remarkable increase in fatigue strength of rotating beam specimens (curve D in Fig. 9). Because of the cyclic stability of the residual stresses the crack nucleation for the shot peened specimens was also below the surface except for the highest stress amplitude in laboratory air. The difference between the curves D and C was due to the reduced applied stress at the depth of crack nucleation but even more to the remarked influence of the stress gradient on propagation on this crack in the rotating beam test (curve D). Therefore a drastic increase in fatigue strength over the electrolytically polished state which exhibited crack nucleation at the surface was obtained by testing in air (compare A and D). On the other hand the fatigue strength of the electrolytically polished condition in vacuum could not be achieved (see difference between curves B and D). This can be due only to the fact that the combination of residual tensile stress and stress gradient led to earlier crack nucleation in the interior of the rotating beam specimens as compared to crack nucleation at the surface in the electrolytically polished condition. For this comparison it should be pointed out that for a given applied stress configuration and crack nucleation in the interior any difference in fatigue life of shot peened specimens tested in vacuum and aggressive environment could only result from the final stage of crack propagation after the crack reaches the surface . No difference in fatigue strength between vacuum, air or 3.5 % NaCℓ solution was therefore observed under these conditions.

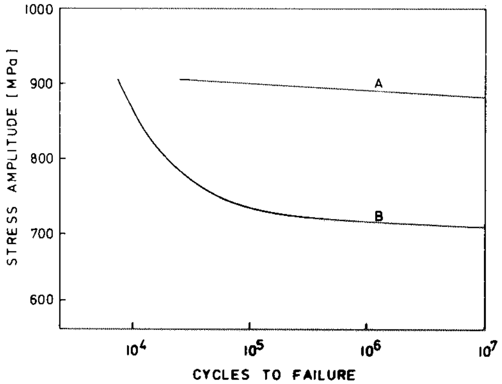

Fig. 5

Alloy condition B/T-RD,
S-N curves for different surface conditions (rotating beam, air)

A: mechanically polished
B: electrolytically polished

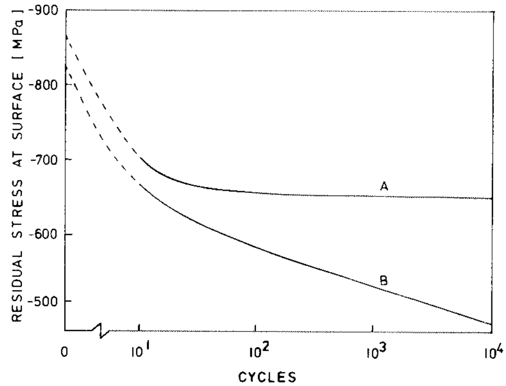

Fig. 6

Residual stress vs. number of cycles (bending, σ_A = 450 MPa, σ_m = - 600 MPa)

A: Alloy condition B/T-TD
B: Alloy condition B/T-RD

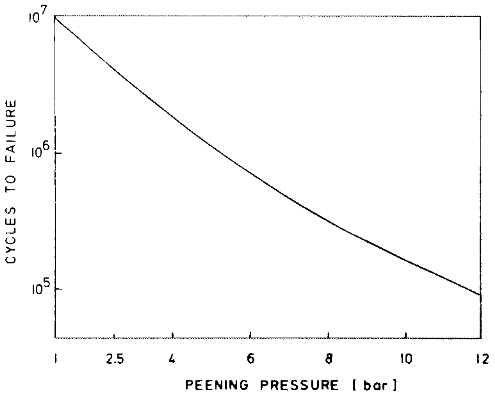

Fig. 7

Alloy condition B/T-TD,
fatigue life vs. peening pressure tested in push-pull in vacuum
(σ_A = 800 MPa, R = - 1)

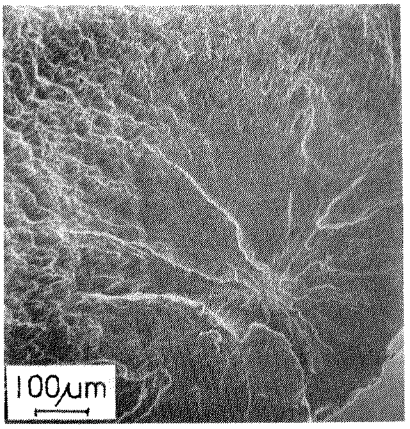

Fig. 8

Alloy condition B/T-TD,
fracture surface (SEM) of a specimen treated 6 bars 4 min.
(see Figure 7)

The corresponding fatigue results of the condtion B/T-RD are given in Fig. 10. Due to the lower cyclic stability of the residual stresses as compared to condition B/T-TD the fatigue cracks for the shot peened specimens nucleated for all stress amplitudes at the surface even in the push-pull tests in vacuum (Fig. 11). In contrast to the push-pull tests of condition B/T-TD (Fig. 9) the shot peening of a material with lower cyclic stability of the residual stresses led to an increase in fatigue strength over the electrolytically polished condition in vacuum (compare curves C and B in Fig. 10). Obviously because of the lower cyclic stability of the residual stresses crack nucleation in the interior was not taking place. The remaining compressive stresses had therefore a remarkable influence on the crack propagation even in the push-pull test because of crack nucleation at the surface. Similar results were obtained by testing in an aggressive environment (compare curves D and A). From the curves it can be seen that crack propagation seems to be more influenced by shot peening than crack nucleation. Only a slight increase in fatigue strength was obtained in an aggressive environment. By comparing the influence of environment on the fatigue behavior (compare curves D and A with C and B) it can be said that the dislocation density seems to be more effective in retardation of crack nucleation in an inert environment.

CONCLUSION

It could be shown that the cyclic stability of the residual stresses and therefore the fatigue behavior of shot peened specimens was strongly influenced by the fatigue hardening or softening behavior of the material. If a material cyclic hardens the stability of the interior residual tensile peak stresses will lead because of early crack nucleation in these areas to a drastic loss in the push-pull fatigue strength particular in an inert environment. The stress gradient in a rotating beam test resulted in a remarkable increase in fatigue strength compared to push-pull tests independently of the crack nucleation site at or below the surface because the residual compressive stresses now had a pronounced effect on crack propagation. On the contrary the less cyclic stability of the residual stresses in a material that cyclic softens will prevent the early crack nucleation in the regions of residual tensile peak stresses and will lead for both the push-pull and the rotating beam method to a more or less pronounced increase in the fatigue strength.

By a separation of the parameters residual stresses, dislocation density and surface roughness it could be shown that the residual stresses had a marked influence on the fatigue behavior mainly by retarding crack propagation. The crack nucleation site (at the surface or in the interior) depended on many parameters (residual stresses and their stability, surface roughness, dislocation density, stress amplitude and gradient, environment). Only for the case of crack nucleation at the surface the high dislocation density caused a drastic improve in fatigue life by retarding crack nucleation. By contrast an increased surface roughness would have the opposite effect on crack nucleation in this case.

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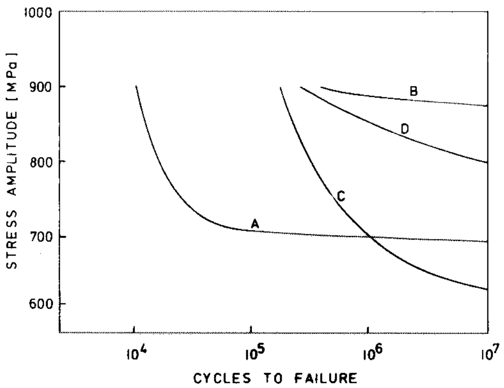

Fig. 9

S-N curves for alloy condition B/T-TD

A: electrolytically polished (air)
B: electrolytically polished (vacuum)
C: shot-peened, push-pull (vacuum)
D: shot-peened, rotating beam (air)

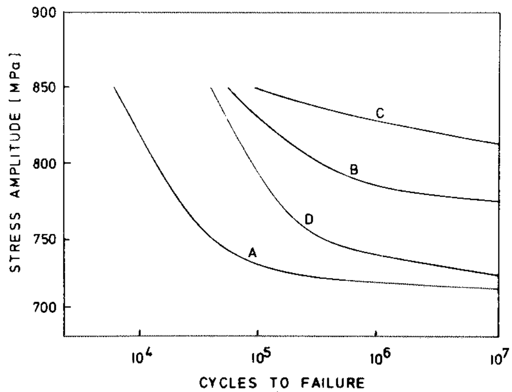

Fig. 10

S-N curves for alloy condition B/T-RD

A: electrolytically polished (air)
B: electrolytically polished (vacuum)
C: shot-peened, push-pull (vacuum)
D: shot-peened, rotating beam (air)

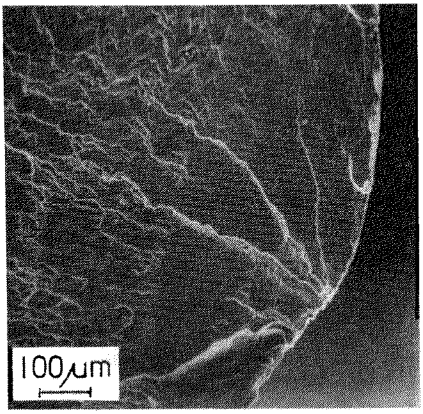

Fig. 11

Alloy condition B/T-RD,
fracture surface (SEM) of a specimen treated
4 bars 4 min., σ_A = 840 MPa
(see Figure 10 C)

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