Influence of Shot Peening Induced Surface Roughness, Residual Macrostresses and Dislocation Density on the Elevated Temperature HCF-Properties of Ti alloys

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Introduction

The fatigue behavior of engineering components is known to be strongly dependent on the properties of the surface and near surface regions. Mechanical surface treatments such as shot peening induce a high dislocation density in near surface regions, residual stresses are developed and the surface topography is changed [1,2]. Without a change in the bulk properties of the material, the fatigue life can significantly increase due to an improved resistance to fatigue crack nucleation and propagation of small surface cracks. Since the shot peening procedure is a fatigue process to the surface layer [3,4] it was argued that microcracks may already nucleate during shot peening [3,5]. Even if cracks are not formed during shot peening the fatigue life of shot peened parts will nevertheless be crack propagation controlled because of the remaining defects, dents and overlaps in the surface layer which lead to early crack nucleation [3,5]. Therefore, the fatigue strength of shot peened specimens should correlate with the threshold value ΔK_{th} for small crack propagation. Previous work [6] has shown that the beneficial effect of shot peening on the fatigue behavior of high-strength titanium alloys at room temperature mainly derives from a marked retardation of the growth rate of small surface cracks propagating within the regions of induced residual compressive stresses. Unlike the compressive residual stresses, the high dislocation densities were found to even accelerate the growth rate of small surface cracks [6]. Obviously, the effect of the last parameter the surface roughness can be very pronounced since the fatigue behavior can change from crack propagation controlled (rough surface condition) to crack nucleation controlled (smooth surface condition) if additional polishing is used after shot peening [3,7,8]. In that case the fatigue strength is more determined by surface layer yield stress and slip length [9].

By separating the individual effects of residual stresses, dislocation density, and surface roughness on fatigue crack nucleation and propagation of small surface cracks at various temperatures the present study was aimed to contribute to the understanding of the mechanisms by which shot peening can improve or worsen the elevated temperature fatigue behavior of titanium alloys.

Experimental Procedure

The investigation was performed on the $(\alpha+\beta)$ Ti-6Al-4V alloy with a fine lamellar (β-quenched) microstructure. The heat treatment and the results of tensile tests at 20°C, 350°C and 500°C (initial strain rate $\dot{\varepsilon}$= 8,4 x $10^{-4}s^{-1}$) are summarized in Table 1.

HEAT TREATMENT	TEST TEMPERATURE	E (GPa)	$\sigma_{0.2}$ (MPa)	$\varepsilon_F = \ln A_O/A_F$
15' 1050 °C/WQ	20 °C	120	1040	0.20
1h 800 °C/WQ	350 °C	104	620	0.30
24h 600 °C/AC	500 °C	96	535	0.53

Table 1: Heat treatment and tensile properties of the fine lamellar microstructure of Ti-6Al-4V

After machining and mechanical polishing, electrolytical polishing was used to remove a layer of about 100 μm from the surface of the fatigue specimens to ensure that any machining effect which could mask the results was absent. This electrolytically polished condition was taken as the reference to which the results of shot peened specimens will be compared.

Shot peening was done using S 110 steel shot and an Almen intensity of I = 0.28 A [mm]. Details of this treatment are given elsewhere [8]. Of these shot peened specimens, some were given an additional stress relief treatment of 1h at 600 °C. In addition, of both the shot peened as well as the shot peened and stress relieved specimens, some were electrolytically repolished. By this means, a surface layer of about 50 μm was removed which resulted in the same smooth surface finish as present in the electrolytically polished reference condition.

The effectiveness of the stress relief treatment was proved by comparing the residual stress distribution as obtained from x-ray measurements and calculated by the $\sin^2\psi$-method [10] before and after the heat treatment. The change in dislocation density due to the annealing treatment was studied by TEM. The surface roughness was measured by a profilometer.

Fatigue tests were performed on smooth hour-glass shaped specimens (gauge diameter: 3.6 mm) in rotating beam loading (R = -1). The tests were done at 20 °C, 350 °C and 500 °C in air at a frequency of about 50 Hz. The specimens were heated by an induction coil driven by an HF-generator, and the temperature was controlled by a calibrated pyrometer.
The propagation of small surface cracks in the baseline and in the shot peened conditions was studied by cycling at a constant stress amplitude of σ_a = 775 MPa for the tests at 20 °C. In order to propagate cracks from the surface of shot peened specimens instead of nucleating new cracks below the surface, first a stress amplitude of σ_a = 850 MPa was applied for about 5000 cycles in the test at 20 °C before changing to the stress amplitude of σ_a = 775 MPa. Constant stress amplitudes of 450 and 400 MPa were used in the elevated temperature tests at 350 and 500 °C, respectively. To monitor surface crack length, the tests were interrupted after defined load cycles and the specimen surface was studied by LM. To enable a microscopic study of microcracks in the shot peened conditions, these specimens were only lightly electrolytically repolished (10 μm removal from the surface) and

etched. During polishing, care was taken not to remove all overlaps and dents in the rough as peened surface because these defects easily served as "crack starters".
Since it was expected that the residual stress field could characteristically change the semicircular "equilibrium" crack depth profile of small surface cracks, the depth profiles of individual surface cracks were determined by a stepwise surface layer removal through electropolishing. The ΔK-values of the surface cracks were calculated for the crack tip in the interior of the specimens applying the equation

$$\Delta K = \Delta\sigma \sqrt{\pi a}\ y/\emptyset \qquad (1)$$

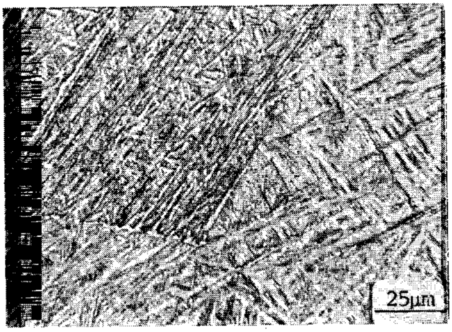
(where: ΔK amplitude of stress intensity factor, $\Delta\sigma$ stress amplitude, a crack depth, y correction factor for crack shape, and $\emptyset$ correction factor for crack depth/ specimen thickness ratio) [11]. Since the specimens were fatigued in fully reversed loading ($R = -1$) only the positive stress range was taken for calculating ΔK.

Experimental Results

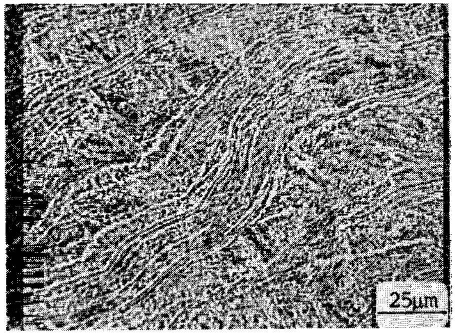
The fine lamellar microstructure of the Ti-6Al-4V alloy is shown in Figure 1. Compared to the baseline condition (Figure 1a), shot peening resulted in marked deformation of the formerly straight lamellae in the near surface region, which now are heavily curved and bent (Figure 1b, compare Figure 1b with Figure 1a).
A TEM-analysis showed that the low dislocation density of the baseline condition (Figure 1c) drastically increased after shot peening. Even after the stress relief treatment of 1h at 600°C, a high dislocation was still present within the surface layer (Figure 1d). By increasing the annealing temperature still further, it was found that the high dislocation density did not significantly decrease until recrystallization took place, which resulted in a fine equiaxed microstructure in the surface layer [12].

The thermal stability of the shot peening induced residual compressive stress profile in the surface layer is shown in Figure 2. Annealing 30 h at 350°C decreased the residual compressive stress at the surface from about 600 MPa to 270 MPa and the maximum value in the interior from 800 MPa to 600 MPa.
Annealing at 500°C resulted in an almost total stress relief at the surface after only 1 hour exposure, whereas deeper in the interior residual compressive stresses as high as 400 MPa were still present. After prolonging the holding time at 500°C from 1 hour to 30 hours the maximum compressive stress in the interior decreased to about 200 MPa. An almost total stress relief in the whole specimen was measured after annealing 1h at 600°C, which is also reported in the literature [13,14].

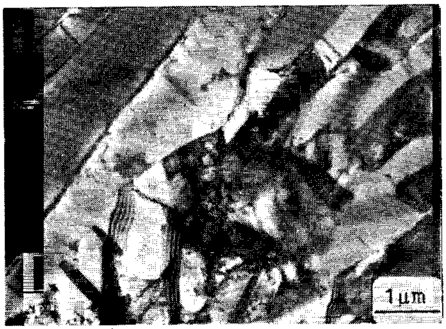
The measured roughness values of the electrolytically polished reference condition were $R_a = 0.07$ µm and $R_t = 0.42$ µm. For the shot peened specimens the corresponding values were 0.58 µm and 5.2 µm, respectively. After electrolytical repolishing the shot



a) electrolytically polished baseline condition (LM)



b) shot peened condition, distance about 20µm from shot peened surface LM



c) baseline condition, low dislocation density (TEM)

d) shot peened and stress relieved 1h 600°C high dislocation density (distance about 20µm from shot peened surface (TEM)

Figure 1: Fine lamellar microstructures of the Ti-6Al-4V alloy

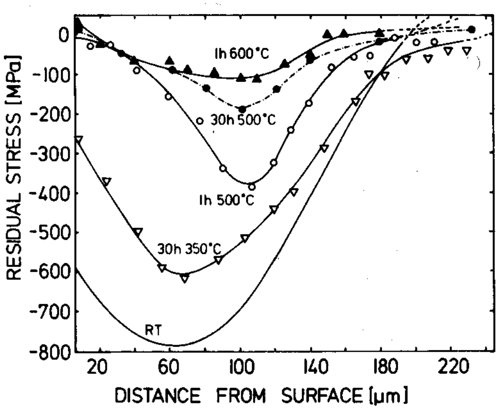
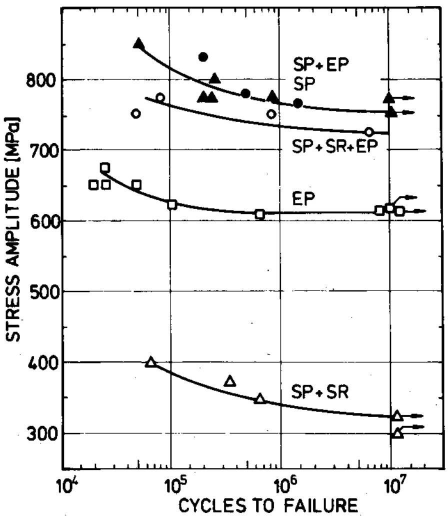
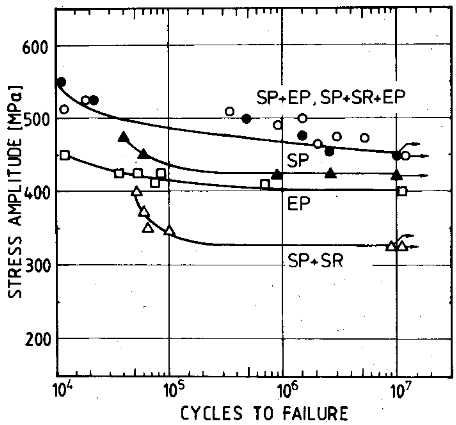


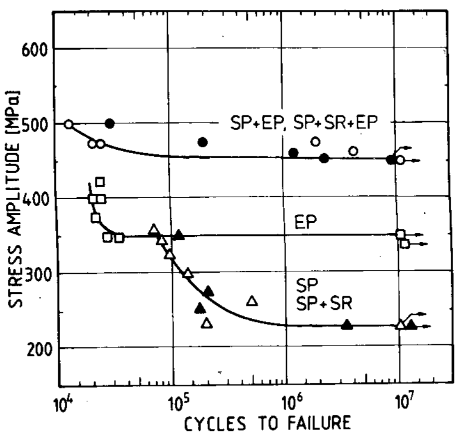
Figure 2: Change of the residual stress profile in the shot peened surface layer of Ti-6Al-4V after various annealing treatments



a) T = 20°C



b) T = 350°C



c) T = 500°C

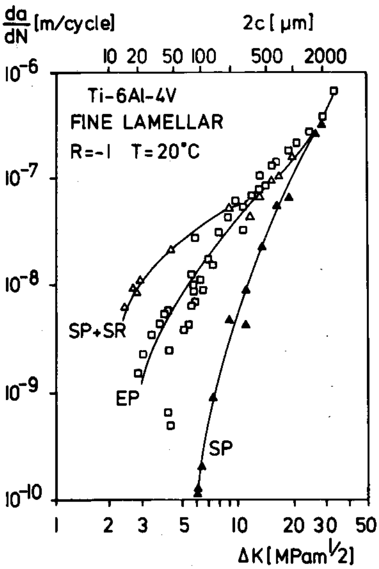
-□- Electrolytically polished (EP)
-▲- Shot peened (SP)
-●- Shot peened and repolished (SP+EP)
-△- Shot peened and stress relieved (SP+SR)
-○- Shot peened, stress relieved and repolished (SP+SR+EP)

Figure 3: S-N curves, Ti-6Al-4V, fine lamellar, 24h 600°C, rotating beam loading (R = -1), air

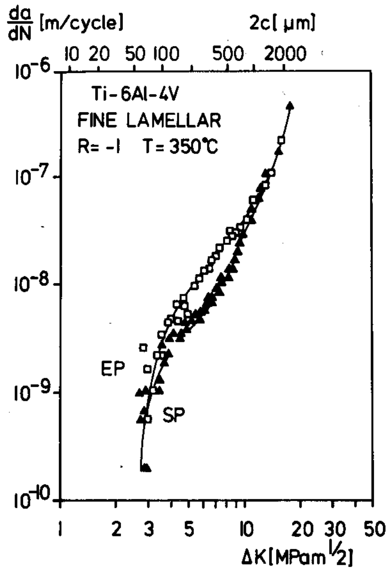
peened specimens and removing about 50 µm from the surface the average measured roughness values were as low as for the reference condition.

The S-N curves obtained in air at 20°C, 350°C, and 500°C are shown in Figure 3a, b, and c, respectively comparing the behavior of the electrolytically polished baseline condition (EP) with the shot peened (SP), shot peened and polished (SP+EP), shot peened and stress relieved (SP+SR), and finally, the shot peened, stress relieved, and polished (SP+SR+EP) condition.
At room temperature the 10^7 cycles fatigue strength increased from about 620 MPa to 760 MPa after shot peening (Figure 3a, compare curve EP with curve SP). Repolishing of shot peened specimens was not found to change the fatigue strength (Figure 3a, compare data points SP with data SP+EP). After the stress relief treatment of 1h at 600°C the fatigue strength of the shot peened condition drastically dropped to about 325 MPa whereas repolishing shot peened and stress relieved specimens led to a marked increase in fatigue strength from 325 up to 720 MPa (compare curve SP+SR with curve SP+SR+EP).
At 350°C, there was only a slight increase of the fatigue strength from 400 MPa to 425 MPa due to shot peening (Figure 3b, compare curve SP with curve EP). The stress relief treatment led to a drop in fatigue strength of shot peened specimens from 425 MPa to 325 MPa (compare curve SP+SR with curve SP). Again, repolishing shot peened specimens led to a marked increase in fatigue strength and to the same high value whether the stress relief treatment was done before or not (compare curve SP and curve SP+SR with curve SP+EP and curve SP+SR+EP, respectively).
At the higher temperature of T = 500°C, shot peening resulted in a marked loss of the fatigue strength from 350 MPa to 225 MPa (Figure 3c, compare curve EP with curve SP). No effect of the stress relief treatment was found, whereas repolishing led to a marked increase of the fatigue strength from 225 MPa to about 450 MPa (compare curve SP and curve SP+SR with curve SP+EP and curve SP+SR+EP, respectively).

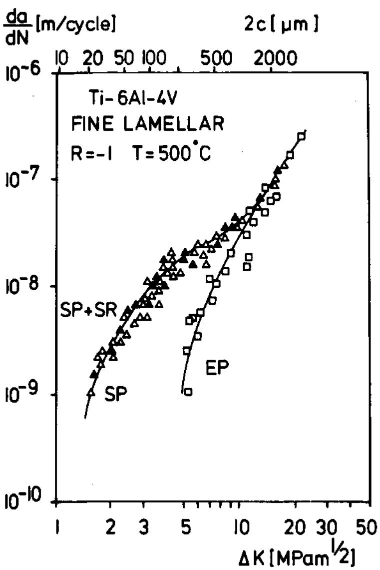
The growth rate of small surface cracks in the baseline and the shot peened conditions for the tests at 20°C, 350°C, and 500°C can be seen in Figure 4a, b, and c, respectively. In addition, for the tests at 20°C and 500°C the growth rate of the shot peened and stress relieved condition is plotted (Figure 4a and c).
Small cracks in the shot peened specimens tested at 20°C (Figure 4a) propagate at a much lower rate, e.g. at a ΔK-value of 6 MPa $m^{1/2}$ the growth rate is more than two orders of magnitude lower as compared to the shot peened and stress relieved condition (compare curve SP with curve SP+SR). It can be seen that the threshold for small fatigue cracks is by far the highest in the shot peened condition. At higher ΔK-values (i.e. longer crack lengths, 2c > 300 µm) the crack propagation rates of the electrolytically polished, the shot peened, and the shot peened and stress relieved condition converge (Figure 4a).
At a test temperature of 350°C, the growth rate of small fatigue cracks in shot peened specimens was on average still somewhat lower as compared to the baseline condition (Figure 4b, compare



a) T = 20°C
σ_a = 775 MPa



b) T = 350°C
σ_a = 450 MPa



c) T = 500°C
σ_a = 400 MPa

-□- Electrolytically polished (EP)
-▲- Shot peened (SP)
-Δ- Shot peened and stress relieved (SP+SR)

Figure 4: Fatigue crack propagation of small surface cracks in Ti-6Al-4V at various test temperatures in air

curve EP with curve SP).
However, at 500°C, small fatigue cracks were found to propagate much faster in the shot peened as compared to the baseline condition (Figure 4c, compare curve SP with curve EP). The stress relief treatment of 1h at 600°C was not seen to further change the growth rate of small cracks in shot peened specimens (Figure 4c, compare curve SP with curve SP+SR).

Discussion

Because the influence of each of the main parameters high dislocation density, residual stresses, and surface roughness on the fatigue strength of shot peened specimens can be quite different depending on temperature, the fatigue results at the various temperatures will be discussed in terms of the actual contribution of each of these parameters. The individual effects will be interpreted with respect to fatigue crack nucleation and propagation of small surface cracks. Since at a rough surface fatigue cracks nucleate early in life, the fatigue behavior will be mainly crack propagation controlled for the conditions SP and SP+SR (rough surface) and crack nucleation controlled for the conditions with a smooth surface (EP, SP+EP, SP+SR+EP).
Therefore, the conditions having a rough surface (SP, SP+SR) will be discussed separately from the conditions having a smooth surface (EP, SP+EP).

1. Rough surface condition (fatigue strength crack propagation controlled).
The resistance to the propagation of small surface cracks at the various test temperatures of the conditions SP and SP+SR can be evaluated from the da/dN-ΔK curves in Figure 4. The threshold values ΔK_{th} for small surface cracks in the shot peened conditions (SP), were about 6, 3, and less than 2 MPa $m^{1/2}$ at 20°C, 350°C, and 500°C, respectively. A comparison of these threshold values with the measured 10^7 cycles fatigue strength values of σ_{a10^7} = 760, 425, and 225 MPa at 20°C, 350°C, and 500°C, respectively (compare curves SP in Figure 4a, b, and c with curves SP in Figure 3a, b, and c) indicates that the fatigue behavior in fact is crack propagation controlled.
The same trend can be found for the shot peened and stress relieved (SP+SR) condition by comparing the 10^7 cycles fatigue strength values of σ_a = 325 and 225 MPa at 20°C and 500°C, respectively with the corresponding threshold values for small crack propagation ΔK_{th} of about 2.5 and less than 2 MPa $m^{1/2}$ at 20°C and 500°C, respectively (compare curves SP+SR in Figure 3a and c with curves SP+SR in Figure 4a and c).
The difference in the fatigue strength of the shot peened (SP) and the shot peened and stress relieved (SP+SR) condition (Figure 3) which is about 435, 135 and 0 MPa at 20°C (Figure 3a), 350°C (Figure 3b), and 500°C (Figure 3c), respectively can be correlated with the effect of residual compressive stresses on small crack growth at these temperatures (compare curves SP with curves SP+SR in Figure 4a and c. This effect is explained in detail in [8]. With an increase in the test temperature the observed decreasing contribution of residual compressive stresses to the

propagation of small surface cracks and thus on the fatigue strength can be explained by the temperature dependent decay of the residual compressive stress in the surface layer (Figure 2).

Whereas the residual compressive stresses retard crack propagation, the high dislocation density was found to accelerate crack growth, a result also observed on long through-cracks in predeformed material [15]. The negative effect of the high dislocation density on small crack propagation can be seen in Figure 4a and c by comparing the curves SP+SR (high dislocation density) with curves EP (low dislocation density). The higher growth rate and lower threshold values in the shot peened and stress relieved conditions can be attributed to the largely reduced ductility in the surface layer of these specimens [6].

2. Smooth surface condition (fatigue strength crack nucleation controlled):
The HCF-properties of specimens having no stress concentrations at the surface is usually stress controlled since the fatigue cracks nucleate late in life, i.e. the crack propagation is only a small portion of the total life [16]. Thus, the ranking of the fatigue strength for different conditions has nothing to do with the ΔK_{th}-values of small surface cracks in these conditions and can even be opposite to the ranking of ΔK_{th}. For example, the 10^7 cycles fatigue strength of the electrolytically polished baseline (EP) exhibited a pronounced drop with test temperature from 620 MPa at 20°C to 350 MPa at 500°C (compare curves EP in Figure 3a and c) while ΔK_{th} increased with temperature from about 3 MPa $m^{1/2}$ at 20°C to 5 MPa $m^{1/2}$ at 500°C (compare curves EP in Figure 4a and c). However, from Table 1 it can be seen that the fatigue strength can be well correlated with the measured yield stress-values at these temperatures. For titanium alloys which do not contain any inclusions or other defects this relationship between fatigue strength and yield stress usually holds [17].

If the surface layer of fatigue specimens is strengthened by a high dislocation density (increase in yield stress) the fatigue strength should also increase as long as no cracks are induced and the surface topography is kept smoothly. The increase in the 10^7 cycles fatigue strength from the baseline to the shot peened, stress-relieved and repolished specimens at the various test temperatures (compare curves SP+SR+EP with curves EP in Figure 3a, b, and c) can be explained by an improved resistance to fatigue crack nucleation due to the high dislocation density in the surface layer resulting in an increase in yield stress in these region. Comparing now the curves SP+EP with curves SP+SR+EP in Figure 3a, b, and c, it can be seen that the effect of residual compressive stresses on fatigue crack nucleation (smooth surface) is not so pronounced.

From the foregoing it is obvious that the effect of the surface roughness on the fatigue strength can be very pronounced since depending on the actual surface topography the fatigue strength can be crack propagation controlled (conditions SP, SP+SR, rough surface) or crack nucleation controlled (conditions EP, SP+EP, SP+SR+EP, smooth surface). Without the additional contribution of

residual compressive stresses the change in fatigue strength due to the surface roughness can be as high as 400 MPa at 20°C, 135 MPa at 350°C, and 225 MPa at 500°C (compare curves SP+SR+EP with curves SP+SR in Figure 3).
If the residual compressive stresses are not relieved before testing the corresponding change in fatigue strength was not significant at 20°C, about 60 MPa at 350°C and 225 MPa at 500°C (compare curves SP+EP with curves SP in Figure 3a, b, and c). This increasing contribution of the surface roughness to the fatigue strength with increasing test temperature can be attributed to the temperature dependent decay of the residual compressive stresses (Figure 2).

Summary

Fatigue crack nucleation in shot peened specimens tested at 20°C, 350°C and 500°C is early in life due to the high surface roughness. If the shot peened surface is polished the high dislocation density can retard crack nucleation at all these temperatures. The effect of residual stresses is not so significant.

Fatigue crack propagation is affected by both residual stresses and high dislocation densities. At 20°C, the negative effect of the high dislocation density is strongly overcompensated by the beneficial influence of the residual compressive stresses, much less at 350°C and not at all at 500°C due to the decay of residual stresses.

The fatigue behavior of shot peened specimens is crack propagation controlled unless the rough as peened surface is removed by additional polishing. With increasing application temperatures this additional polishing treatment is becoming more effective in improving the fatigue behavior. At 500°C, shot peening can only improve the fatigue strength if in addition the surface is polished.

Acknowledgements

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References

1. R. Schreiber, H. Wohlfahrt, and E.Macherauch, Arch. Eisenhüttenwes. 48 (1977) 649.
2. H. Wohlfahrt, Shot Peening (edited by A. Niku Lari) Pergamon Press (1982) 675.
3. L. Wagner and G. Lütjering, Shot Peening (edited by A. Niku Lari) Pergamon Press (1982) 453.
4. R.Z. Wang, Y.G. Tan, X.B. Li, M.G. Yan, Shot Peening (edited by A. Niku Lari) Pergamon Press (1982) 185.
5. G.H. Fair, B. Noble, and R.B. Waterhouse, Proc. Int. Conf. on Fatigue of Eng. Mat. and Structures, The Institution of Mechanical Engineers, Sheffield, U.K. (1986) 437.

6. H. Gray, L. Wagner, and G. Lütjering, Proc. of the 1st Int. Conf. on Residual Stresses (edited by E. Macherauch and V. Hauk) Garmisch-Partenkirchen (1986).
7. L. Wagner, C. Gerdes, and G. Lütjering, Titanium, Science and Technology (edited by G. Lütjering, U. Zwicker, and W. Bunk) DGM (1985) 2147.
8. H. Gray, L. Wagner, and G. Lütjering, Fatigue Prevention and Design (edited by J.T. Barnby) Chamelion Press (1986) 363.
9. G. Lütjering, A. Gysler, and L. Wagner, Proc. Advanced Materials Research and Developments for Transport (edited by R.J.H. Wanhill, W.J. Bunk, and J.G. Wurm) Strasbourgh, European Metals Research Society (1985) 309.
10. E. Macherauch and P. Müller, Z. f. angew. Physik (1961) 305.
11. J.C. Newman, ASTM-STP 687 (1979) 16.
12. H. Gray, L. Wagner and G. Lütjering, this conference.
13. Th. Hirsch, O. Vöhringer, and E. Macherauch, HTM 28 (1983) 229.
14. O. Vöhringer, Th. Hirsch, and E. Macherauch, Titanium, Science and Technology (edited by G. Lütjering, U. Zwicker, and W. Bunk) DGM (1985) 2203.
15. K. Schulte, H. Nowack, and G. Lütjering, Engg. Fracture Mech. 13 (1980) 1009.
16. E.A. Starke and G. Lütjering, Fatigue and Microstructure, ASM (1978) 205.
17. G. Lütjering and A. Gysler, Titanium, Science and Technology (edited by G. Lütjering, U. Zwicker, and W. Bunk) DGM (1985) 2065.