2002051

Effect of Shot Peening on Fatigue Performance of Gamma Titanium Aluminides

Janny Lindemann[1], Dan Roth-Fagaraseanu[2] and Lothar Wagner[1]
[1] Chair of Physical Metallurgy and Materials Technology, Technical University of Brandenburg at Cottbus, Cottbus, Germany
[2] Rolls-Royce Deutschland, Dahlewitz, Germany

1 Introduction

Gamma titanium aluminides are attractive candidates for application as blade material in the high-pressure part of the compressor of gas turbines. Compared to the yield stress values at both room temperature and the typical service temperature of 650 °C, the 10^7 cycles fatigue strengths at these temperatures are extraordinarily high, particularly if related to the material's density being only half that of the commonly used Ni-superalloys [1, 2]. To further improve the fatigue performance of γ(TiAl), mechanical surface treatments such as shot peening or roller-burnishing can be utilized. These treatments induce high dislocation densities owing to plastic deformation in near-surface regions, change the surface topography and generate residual compressive stresses. While the effect of process parameters of mechanical surface treatments on fatigue in structural steels, aluminum and conventional titanium alloys was often studied, no such information is available for gamma titanium aluminides. The present work is part of a project which was undertaken to determine potential improvements in room and elevated temperature applications of γ(TiAl) by shot peening.

2 Experimental

The γ(TiAl) base ingot with the chemical composition Ti-47Al-3.7(Nb, Cr, Mn, Si)-0.5B was received from Duriron (USA). The material was hipped, extruded and forged to turbine blades. After forging, the turbine blades were heat treated to achieve fully lamellar microstructures. Specimens for mechanical tests were machined from the turbine blade forgings with the load axis in longitudinal blade direction (L) as indicated in Figure 1.

Tensile tests were performed on electropolished specimens having a gage length of 20 mm and a gage diameter of 4 mm. The initial strain rate was $8.3 \times 10^{-4}\ s^{-1}$. Compression tests were done on cylindrical specimens with a length of 8 mm and a diameter of 4 mm. The initial strain rate was $2.1 \times 10^{-3}\ s^{-1}$. Fatigue tests were performed on hourglass shaped specimens with a gage diameter of 2.5 mm in rotating beam loading ($R = -1$). Some tests were also done on cylindrical circumferentially notched specimens having a notch factor $k_t = 1.7$. The specimen geometries for the various mechanical tests are illustrated in Figure 2.

Shot peening was performed by means of an injector type system using spherical zirconia based ceramic shot with an average diameter of 0.5 mm. The Almen intensity was varied from 0.08 to 0.40 mmN. All peening was done to full coverage.

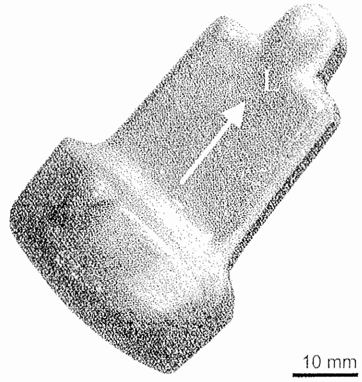

Figure 1: Turbine blade forging

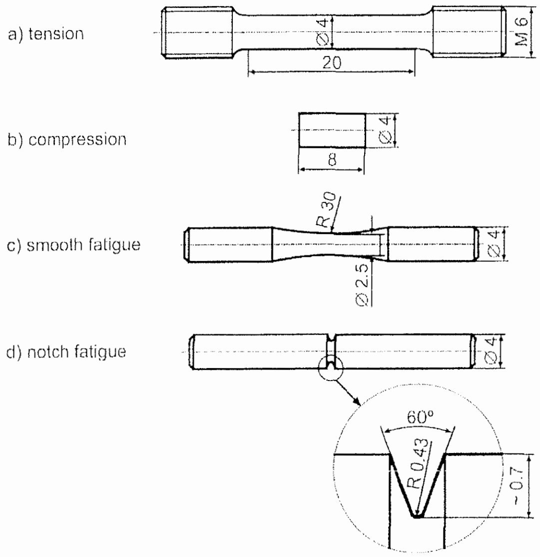

Figure 2: Specimen geometries for mechanical testing

The change in surface layer properties was characterized using profilometry, microhardness and residual stress-depth profile measurements by means of the incremental hole drilling method [3]. The diameter of the drill was 1.7 mm. The oscillating drill was driven by an air turbine with a rotational speed of about 200.000 rpm. The shot peening induced strains in the surface layer were measured with strain gage rosettes at drilled depths of about every 20 µm. The residual stresses at each depth were calculated from the measured strain gage response using a Young's modulus of 170 GPa.

The fatigue results after shot peening were compared with the electrolytically polished reference. At least, 50 µm were removed from the machined surface to make sure that any machining effect that could mask the results was absent.

The thermal stability of the shot peening-induced microhardness and residual stress profiles at an anticipated application temperature of 650 °C was determined by comparing measurements before and after an annealing treatment at this temperature for 50 hours which corresponds to the exposure time for run-outs (10^7 cycles at 60 Hz).

3 Results and Discussion

The typical microstructure within the blade section of the forgings (Fig. 1) is shown in Figure 3. As seen by optical microscopy, the microstructure is fully lamellar with a packet size of about 150–200 µm. The monotonic stress-strain curves of the material are plotted in Figure 4 comparing the differences in mechanical behavior between tensile and compressive loading. The 0.2 % yield stress in compression (580 MPa) is significantly higher than the corresponding value in tension (440 MPa). The apparently lower yield stress in tension might be the result of early crack nucleation and propagation which is also indicated by the low tensile ductility of only about 1 %. On the other hand, the stress-strain relationship in compression illustrates the marked work-hardening capacity of the material which obviously can not be seen in tensile loading due to premature failure.

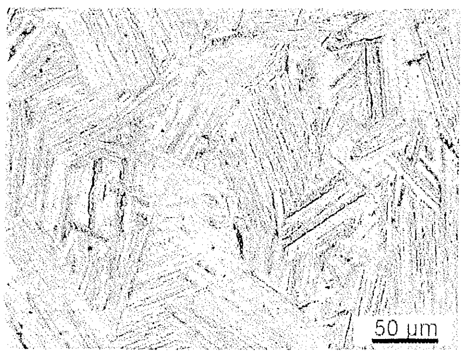

Figure 3: Microstructure of γ(TiAl)

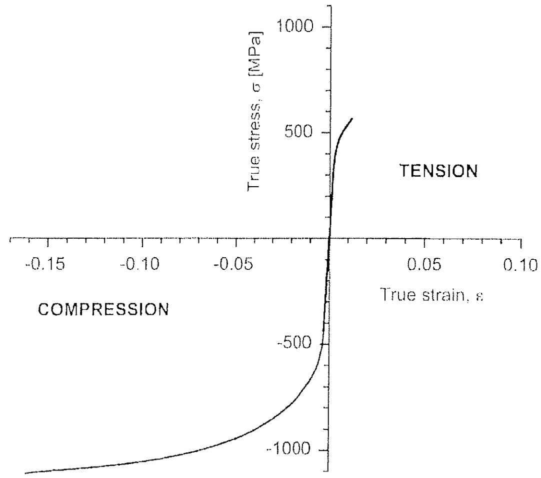

Figure 4: Stress strain curves

The S-N curves of smooth (k_t = 1.0) and notched (k_t = 1.7) specimens of the electropolished reference are shown in Figure 5. Note that the geometrical notch factor of 1.7 reduces the smooth fatigue strength (550 MPa) by only about 10 % to 500 MPa. Thus, the maximum notch root stress $\sigma_a \cdot k_t$ at 10^7 cycles (850 MPa) is by far higher than the smooth (k_t = 1.0) fatigue strength (550 MPa). This result was also found in the literature [4, 5] and is in contrast to previous results on (α+β) titanium alloys, where the geometrical notch factor k_t fully (100 %) affected fatigue strength.

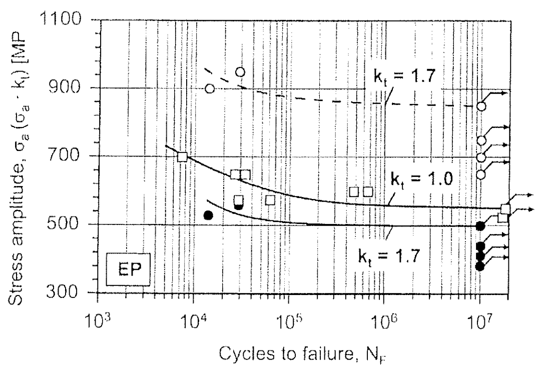

Figure 5: S-N curves in rotating beam loading ($R = -1$), EP

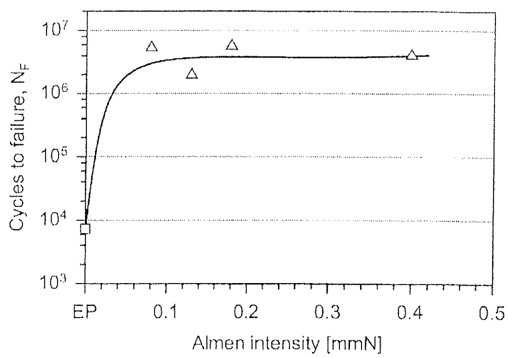

Figure 6: Fatigue life (σ_a = 700 MPa) vs. Almen intensity

The effect of Almen intensity on the fatigue life of smooth (k_t = 1.0) specimens in rotating beam loading ($R = -1$) at a stress amplitude of 700 MPa is plotted in Fig. 6. Starting with the electrolytically polished reference, the fatigue life drastically increases by more than two orders of magnitude. No overpeening effect was found within the range of Almen intensities utilized. Further fatigue testing was done only with specimens shot peened to an Almen intensity of 0.40 mmN.

Shot peening clearly increases surface roughness compared to the electrolytically polished reference (Fig. 7). Due to the process-induced plastic deformation, the microhardness in near-surface regions significantly increases to a maximum value at the surface which is higher than twice the value of the unaffected material in the bulk (Fig. 8). This again illustrates the extraordinarily high work-hardening capacity of the γ(TiAl) material (compare Fig. 8 with Fig. 4). For the given Almen intensity of 0.40 mmN, the process-induced depth of plastic deformation is roughly 300 µm. The residual stress-depth profile as measured by the incremental hole drilling method is plotted in Figure 9 indicating residual compressive stresses in near-surface regions. The maximum value of about –800 MPa was found at the surface followed by a gradual decrease within the plastically deformed surface layer (compare Fig. 9 with Fig. 8).

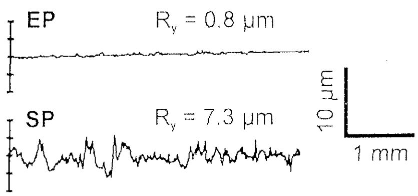

Figure 7: Surface roughness profiles (EP-electropolished, SP-shot peened, 0.40 mmN)

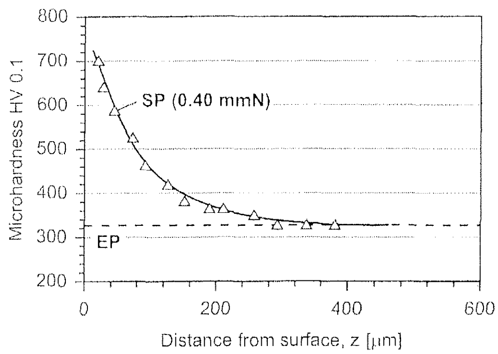

Figure 8: Microhardness profile after shot peening (0.40 mmN)

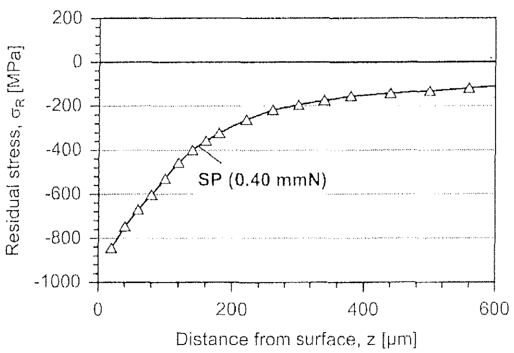

Figure 9: Residual stress profile after shot peening (0.40 mmN)

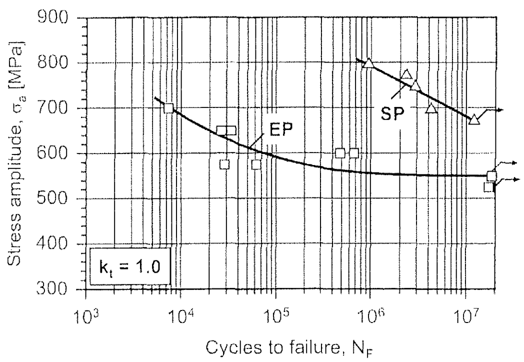

Figure 9: Residual stress profile after shot peening (0.40 mmN)

The resulting S-N curve for smooth (k_t = 1.0) specimens is shown in Figure 10 comparing the fatigue performance between this shot peened condition and the electropolished reference. As seen in Figure 10, the 10^7 cycles fatigue strength of the electropolished condition (550 MPa) increases after shot peening to 675 MPa. This fatigue strength improvement by roughly 23 % is significantly higher than those previously determined on shot peened (α+β) titanium alloys such as Ti-6Al-4V [6, 7] and near-α titanium alloys such as TIMETAL 1100 [8]. Presumably, the shot peening-induced residual compressive stresses in γ(TiAl) are cyclically quite stable and thus, can significantly suppress microcrack growth from the surface to the specimen interior.

As seen in Figure 11, the fatigue crack nucleation site shifts from the surface of the specimens to regions below the surface after shot peening. This shift in crack nucleation site in shot peened specimens is frequently found and is the result of an effective suppression of microcrack growth from the surface to the interior by residual compressive stresses. The resistance to subsurface fatigue crack nucleation depends on the amount and cyclic stability of residual tensile stresses which balance the outer compressive stress field, the mean stress sensitivity of the material and the materials fatigue strength in vacuum.

The S-N curves of the notched (k_t = 1.7) specimens comparing the shot peened with the electropolished condition are shown in Figure 12. The fatigue strength increases from 500 to 630 MPa, i.e., by roughly 26 %, a value only marginally greater than observed on smooth specimens (23 %). Again, this result indicates little notch sensitivity of the HCF strength of γ(TiAl).

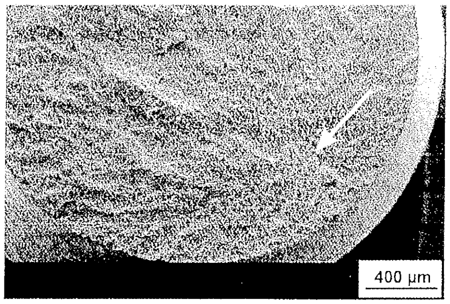

Figure 11: Fatigue crack nucleation site in smooth shot peened specimens

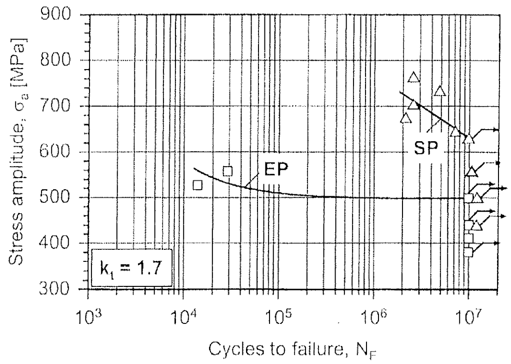

Figure 12: S-N curves (R = −1) of notched specimens

Although no elevated temperature fatigue tests were done in this investigation, a few critical tests were performed which may shed some light on the material's response to shot peening in fatigue at 650 °C. The effect of an annealing treatment at 650 °C for 50 hours on the shot peening induced microhardness in the surface layer is shown in Fig. 13.

Compared to the as-peened condition, there is only a slight decrease in microhardness which indicates that the shot peening-induced high dislocation density is quite stable at 650 °C. On the contrary, the process-induced residual compressive stresses exhibit a marked decay by the same treatment (Fig. 14). Obviously, creep deformation at 650 °C can transform most of the near-surface elastic strains to plastic strains, thus leading to a marked residual stress relief.

Not surprisingly, this annealing treatment at 650 °C for 50 hours significantly deteriorates fatigue performance of shot peened specimens as seen in Figs. 15 and 16. Since both conditions SP and SP + A have a rough surface which contains microcracks, the fatigue strengths are crack propagation controlled. Owing to the residual stress decay in the annealed condition, the threshold value for microcrack growth ΔK_{th} decreases which reduces the fatigue strength particularly, for notched specimens (Fig. 16).

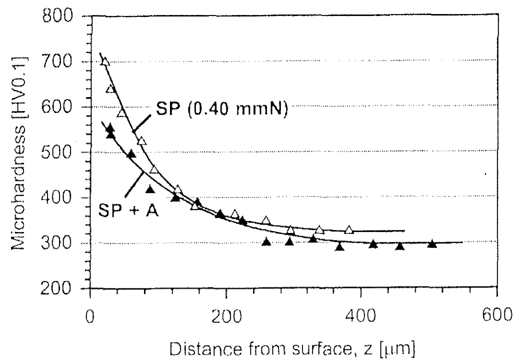

Figure 13: Microhardness profile (0.40 mmN), effect of annealing

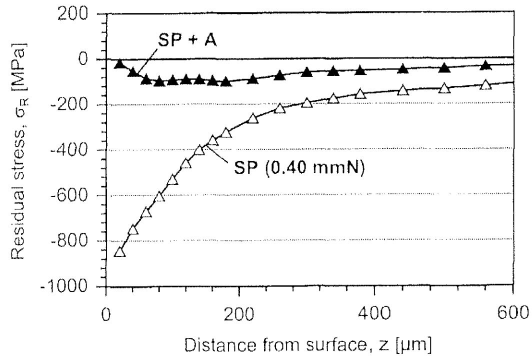

Figure 14: Residual stress profile (0.40 mmN) effect of annealing

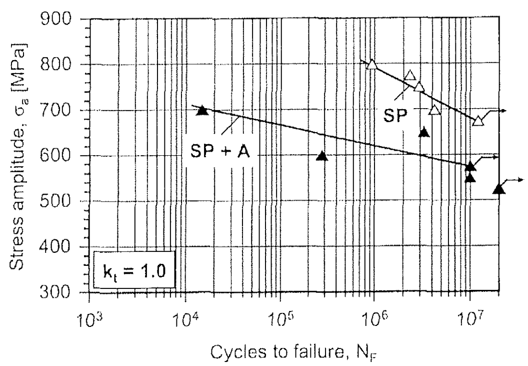

Figure 15: S-N curves (R = -1) of smooth shot peened specimens, effect of annealing

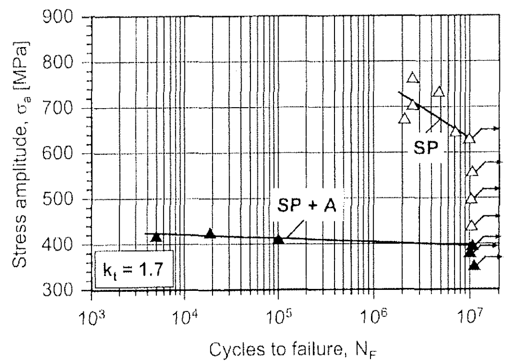

Figure 16: S-N curves (R = -1) of notched shot peened specimens, effect of annealing

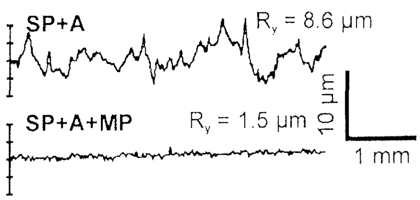

Figure 17: Roughness of shot peened and annealed conditions, effect of mechanical polishing (15 μm removed from as-peened surface)

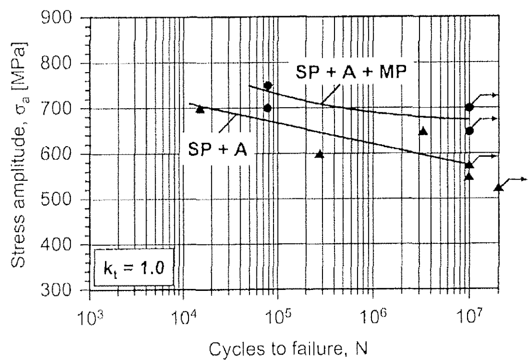

Figure 18: S-N curves (R = −1) of shot peened and annealed conditions, effect of mechanical polishing (15μm removed from as-peened surface)

However, polishing shot peened and subsequently annealed specimens can not only significantly reduce surface roughness (Fig. 17) but also markedly improve fatigue performance as seen in Figure 18. For a smooth surface, the fatigue strength is crack nucleation controlled.

Therefore, the work-hardened surface layer which is still present after the anneal (see Fig. 13) can increase the resistance to fatigue crack nucleation and thus, improve the fatigue strength.

Regarding fatigue performance of shot peened γ(TiAl) specimens at T = 650 °C, it may be derived from the annealing experiments that the work-hardened surface layer will be beneficial for improving elevated temperature fatigue strength. However, the peened surface will need polishing to take advantage of the beneficial effect of the high strength of the surface layer on fatigue crack nucleation.

4 References

[1] W. E. Dowling et al., Microstructure/Property Relationships in Titanium Aluminides and Alloys (Eds.: Y.-W. Kim and R. R. Boyer) TMS, Warrendale, PA, 1991, 123.

[2] J. Kumpfert, Y.-W. Kim, and D. M. Dimiduk, Mat. Sci. and Eng., A192/193, 1995, 465.

[3] T. Schwarz, H. Kockelmann, VDI Report 940, 1992, 99.

[4] S. J. Trail and P. Bowen, Gamma Titanium Aluminides (Eds.: Y. W. Kim, R. Wagner and M. Yamaguchi) TMS, 1995, 883.

[5] W. V. Vaidya, K.-H. Schwalbe and R. Wagner, Gamma Titanium Aluminides (Eds.: Y. W. Kim, R. Wagner and M. Yamaguchi) TMS, 1995, 867.

[6] L. Wagner and G. Lütjering, Shot Peening (Ed.: H. O. Fuchs) American Shot Peening Society, 1984, 194.

[7] L. Wagner, Fatigue Behavior of Titanium Alloys, (Eds.: R. R. Boyer, D. Eylon and G. Lütjering) TMS, 1999, 253.

[8] T. Dörr and L. Wagner, Surface Performance of Titanium Alloys (Eds.: J. K. Gregory, H. J. Rack and D. Eylon) TMS-AIME, 1996, 231.