Effect of Modified Surface Layer Microstructures through Shot Peening and Subsequent Heat Treatment on the Elevated Temperature Fatigue Behavior of Ti Alloys

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Introduction

Shot peening is well recognized as a mechanical surface treatment which can significantly improve the fatigue resistance of high-strength materials [1,2]. The potential application of shot peening is often thought to be necessarily restricted to low temperatures since the residual compressive stresses which are known to be of most importance in improving the fatigue strength at low temperatures [3-7] can readily relax at elevated temperatures due to recovery of the material [8,9]. However, previous work on Ti-alloys [10,11] has demonstrated that the 10^7 cycles fatigue strength at elevated temperatures can be significantly increased without the presence of residual compressive stresses due to the thermal stability of the induced high dislocation density. To take advantage of the beneficial effect of high dislocation densities on fatigue crack nucleation the rough as peened surface was smoothened by additional polishing treatments [10-12].

Lately, it was observed that smaller grains or finer phase dimensions in Ti-alloys besides the beneficial effect on fatigue crack nucleation [13-15] can also improve the resistance to propagation of small surface cracks [16-18]. In the recent past, developments of high temperature titanium alloys have reflected the increased drive for optimizing their all-round capabilities [19]. While it seems to be quite difficult to optimize a microstructure with regard to the resistance against creep and fatigue [15,20] shot peening may offer a method for combining excellent fatigue with superior creep properties in engineering components, such as gas turbine blades. By shot peening of an engineering component with a coarse creep resistant microstructure and by subsequent annealing that component at sufficiently high temperatures, very fine recrystallized (fatigue resistant) phase dimensions can be produced in near surface regions [21] without changing the coarse (creep resistant) microstructure in deeper regions.

In the present investigation, shot peening and subsequent recrystallization was applied on fatigue specimens of Ti-alloys to modify the surface layer microstructures for an improved resistance to elevated temperature fatigue behavior.

Experimental Procedure

The investigation was performed on a binary Ti-alloy with a composition (in wt.%) of 8.6 Al and 0.1 O_2 and on the near-α Ti-6242 (6.0 Al, 2.0 Sn, 4.0 Zr, 2.0 Mo, 0.09 Si, 0.11 O_2). After the homogenization treatment of 1 h at 950°C/WQ, the binary alloy was unidirectionally hot rolled at 950°C (ϕ = -1.4/WQ) and recrystallized at 950°C for 24 hours to obtain an equiaxed large grain size. The Ti-6242 alloy was homogenized 0.5 h at 1050°C and then furnace cooled from the β-field with a cooling rate of 37°C/min to 850°C followed by air cooling.

From these materials fatigue specimens were machined. A few specimens were directly final heat treated (Ti-8.6Al: 10h at 500°C; Ti-6242: 2h at 650°C) and subsequently electrolytically polished by which a layer of about 100 μm was removed from the surface. A few specimens were first shot peened after machining and then recrystallization annealed. For the binary Ti-8.6Al alloy steel shot S 230 was used with an Almen intensity of I = 0.28 A [mm], while for Ti-6242 because of its much higher strength steel shot S 550 was used with an Almen intensity of I = 0.61 A [mm]. After shot peening a recrystallization treatment of 1 h at 820°C was done for the Ti-8.6 Al alloy and of 1 h at 850°C for the Ti-6242 alloy to produce the desired refinement in microstructures. After the recrystallization treatment the specimens received the same final heat treatment as the reference condition. Before fatigue testing the shot peened and recrystallized specimens, a layer of about 50 μm was removed from the surface by electrolytical repolishing which resulted in the same smooth surface finish as present in the baseline (reference) condition.

To evaluate the change in mechanical properties due to the microstructural refinement, tensile tests were done comparing the starting baseline material with further thermomechanically treated materials. The thermomechanical treatment was performed to produce specimens having bulk properties equivalent to those present in the surface layer of shot peened and recrystallized fatigue specimens. In order to develop a similar texture in the tensile specimens as present in the surface layer of the fatigue specimens which is presumably of the basal-type because of the rotation-symmetrical plastic deformation normal to the shot impact, cross rolling (CR) was performed for the Ti-6242 alloys while for the Ti-8.6Al alloy unidirectional rolling (UR) could be used [22,23]. The heat treatment and the results of tensile tests (initial strain rate $\dot{e} = 8.4 \times 10^{-4} s^{-1}$) are summarized for the Ti-8.6Al alloy and the Ti-6242 alloy in Table 1 and Table 2, respectively.

*HEAT TREATMENT	GRAIN SIZE (μm)	$\sigma_{0.2}$ (MPa)		$\varepsilon_F = \ln A_0/A_F$	
		20°C	350°C	20°C	350°C
24h 950°C/WQ 1h 820°C/WQ	100	750	460	0.08	0.62
1h 820°C/WQ	20	815	545	0.11	0.65

* Prior treatment:
1h 950°C/WQ
UR at 950°C (ϕ = -1.4/WQ)

Final treatment:
10h 500°C

Table 1: Microstructures and mechanical properties of the Ti-8.6Al alloy

*HEAT TREATMENT	MICRO-STRUCTURE	$\sigma_{0.2}$ (MPa) 20 °C	550 °C	$\varepsilon_F = \ln A_0/A_F$ 20 °C	550 °C
1h 850 °C/AC	Fine lamellar	1000	550	0.16	0.22
CR at 850 °C (ϕ=-1.4/AC) 1h 850 °C/AC	Fine equiaxed	985	575	0.28	0.69

* Prior treatment:

0.5h 1050 °C $\xrightarrow{37\,°C/min}$ 850 °C/AC

Final treatment:
2h 650 °C/AC

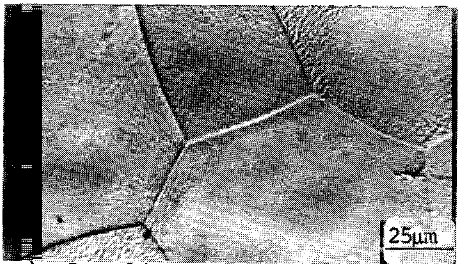
Table 2: Microstructures and mechanical properties of the Ti-6242 alloy

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 350 °C (Ti-8.6Al) and 550 °C (Ti-6242) 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 formation of slip bands and the nucleation of fatigue cracks was studied by optical microscopy.

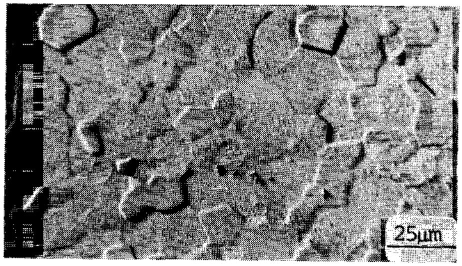
Experimental Results

The microstructure of the binary Ti-8.6Al alloy is shown in Figure 1. Compared to the baseline material (Figure 1a) shot peening and subsequent recrystallization annealing led to a refinement in grain size in near surface regions from about 100 µm to 20 µm (Figure 1b, compare Figure 1b with Figure 1a). From the cross-section of a specimen (Figure 1c) it can be seen that the recrystallized fine grained layer had a thickness of about 150 µm. For the near-α Ti-6242 alloy (Figure 2) shot peening and recrystallization treatment was found to transform the fine lamellar starting microstructure (Figure 2a) into a fine equiaxed microstructure (Figure 2b). The former β grain size in the fine lamellar microstructure (Figure 2a) was about 500 µm and the thickness of the α lamellae about 3 µm. For the fine equiaxed microstructure (Figure 2b) the α-grain size was also about 3 µm. The thickness of the recrystallized fine equiaxed microstructure was about 200 µm (Figure 2c).

The S-N curves in air at 350 °C obtained for the binary Ti-8.6Al alloy are shown in Figure 3. Compared to the starting large grain size material the grain refinement in the near surface region significantly improved the fatigue behavior. The 10^7 cycles fatigue strength increased from 270 MPa up to 320 MPa due to the combination of shot peening and recrystallization annealing. At higher stress amplitudes the fatigue life increased by a



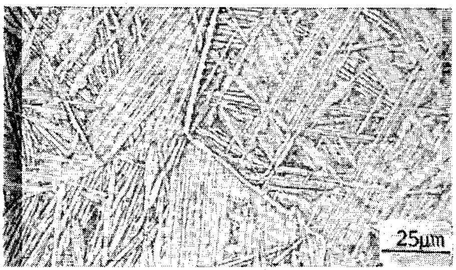
a) Baseline condition



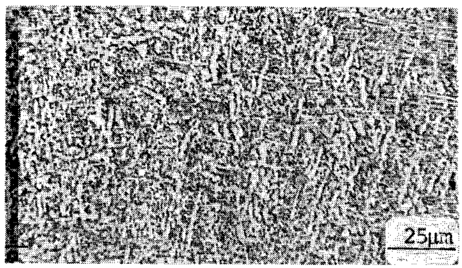
b) Shot peened and recrystallized (distance about 50μm from shot peened surface)

c) Cross-section of shot peened and recrystallized specimen

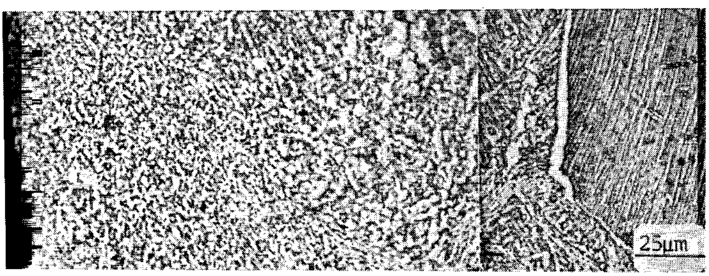
<u>Figure 1:</u> Microstructures of the Ti-8.6Al alloy (LM)



a) Baseline condition



b) Shot peened and recrystallized (distance about 50 μm from shot peened surface)



c) Cross-section of a shot peened and recrystallized specimen

<u>Figure 2:</u> Microstructures of the Ti-6242 alloy (LM)

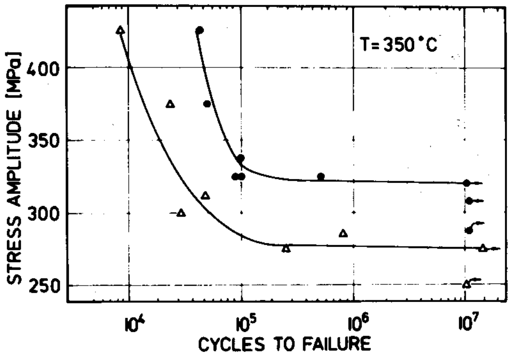
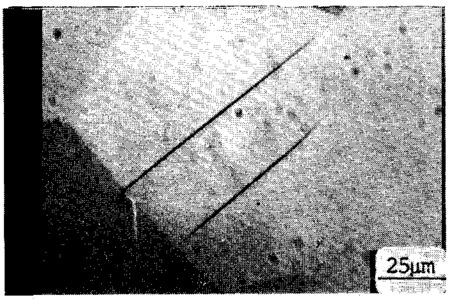
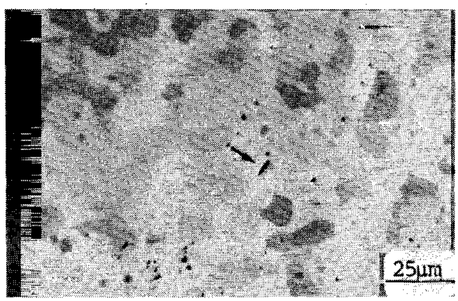


Figure 3: S-N curves, Ti-8.6Al, 10h 500°C
Rotating beam loading (R = -1), air
-Δ- baseline condition (100 μm grain size)
-●- shot peened and recrystallized
(20 μm grain sized surface layer)



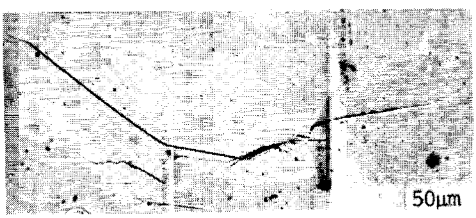


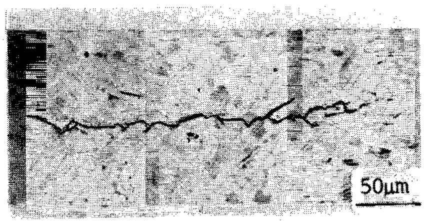
a) Baseline condition
σ_a = 300 MPa
N = 1 x 10^4

L.A.

b) Shot peened and recrystallized, σ_a = 325 MPa
N = 3 x 10^4

Figure 4: Formation of slip bands and fatigue crack nucleation in Ti-8.6Al (compare with Figure 3)





a) Baseline condition
σ_a = 300 MPa
N_F = 3.1 x 10^4

L.A.

b) Shot peened and recrystallized, σ_a = 325 MPa
N_F = 1.1 x 10^5

Figure 5: Small cracks in Ti-8.6Al (compare with Figure 3)

factor of about 5. It was observed that the formation of intense planar slip bands and the subsequent nucleation of fatigue cracks along these slip bands (Figure 4) was earlier in the coarse grain (Figure 4a) as compared to the fine grain material (Figure 4b). For the two different grain sizes typical small surface cracks are shown in Figure 5. It can be seen that for reaching the same surface crack length 2c of about 300 µm the fatigue crack in the fine grain material (Figure 5b) had to change its growth direction much more often as compared to the small crack in the coarse grain material (Figure 5a).

The S-N curves in air at 550°C for the Ti-6242 alloy are shown in Figure 6. Similar as found for the binary Ti-8.6Al alloy, a marked increase of the 10^7 cycles fatigue strength was found after shot peening and recrystallization annealing. As can be seen in Figure 6 the fatigue strength increased from about 390 to 450 MPa. Unlike the situation for the binary Ti-8.6Al alloy the fatigue life improvement at higher stress amplitudes was less significant for the Ti-6242 alloy (compare Figure 6 with Figure 3). While small surface cracks appeared to propagate along the interface between the α and β lamellae in the fine lamellar baseline microstructure (Figure 7a) with a more or less pronounced deviation from the normal to the stress axis, small cracks in the modified fine equiaxed microstructure were remarkably straight and oriented perpendicular to the stress axis (Figure 7b).

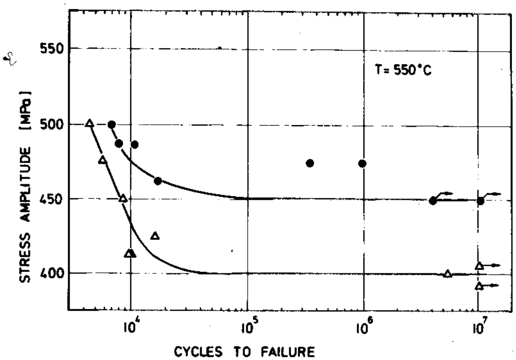
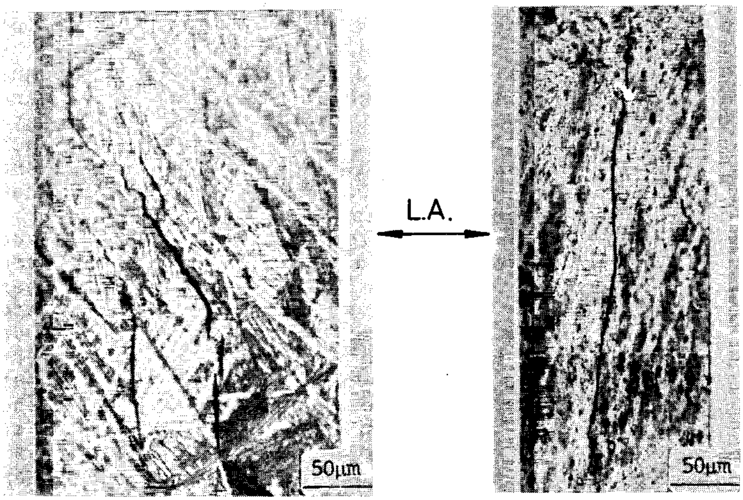


Figure 6: S-N curves, Ti-6242, 2h 650°C
Rotating beam loading (R = -1), air
-Δ- Baseline condition (fine lamellar)
-●- Shot peened and recrystallized (fine equiaxed surface layer)



a) Baseline condition (fine lamellar) σ_a = 475 MPa N_F = 6 x 10^3

b) Shot peened and recrystallized (fine equiaxed surface layer) σ_a = 475 MPa N_F = 3.5 x 10^5

Figure 7: Small cracks in Ti-6242

Discussion

The presented fatigue results on the binary Ti-8.6Al alloy and on the near-α Ti-6242 alloy demonstrate that the elevated temperature fatigue behavior can be improved through shot peening and subsequent recrystallization. This effect is clearly due to the refinement in the surface layer microstructures since both residual compressive stresses and the high dislocation densities which are also known to improve the fatigue strength of high-strength titanium alloys [10-12] were absent. The increase in the 10^7 cycles fatigue strength from 270 to 320 MPa as observed for the binary Ti-8.6Al alloy (Figure 3) can be explained by an increased resistance to fatigue crack nucleation within the surface layer of shot peened and recrystallized specimens. It was found that this grain refinement from 100 μm to 20 μm increased the yield stress at 350°C from 460 to 545 MPa (Table 1). As a result the formation of slip bands and the nucleation of fatigue cracks at these slip bands was observed later in life for the shot peened and recrystallized condition compared to the coarse grained baseline (Figure 4, compare Figure 4a with Figure 4b). On the other hand, the improvement of the fatigue life at higher stress amplitudes may result from an increased resistance to the propagation of small surface cracks in the fine grained surface layer microstructure of the shot peened and recrystallized specimens. From room temperature fatigue testing [17] it is known that a decrease in grain size from 100 to 20 μm decreases the growth rate of small surface cracks in Ti-8.6Al by a factor of about 5. The higher resistance to small crack propagation was explained

primarily with the higher density of grain boundaries in the fine grained material because these grain boundaries were observed to be effective barriers for small cracks [17]. Since in the fatigue tests at 350°C of the present investigation fatigue cracks were also seen to be forced to change direction when propagating into adjacent grains (Figure 5) the higher fatigue life in the shot peened and recrystallized specimens in the LCF-region, where cracks nucleate early in life, can probably as well be explained by an increased resistance to small crack growth.
It should be noted that the nucleation of subsurface fatigue cracks below the fine grained layer in regions of the coarse grained microstructure was hindered not only by the stress gradient in bending but even more by the fact that subsurface cracks have to nucleate under vacuum conditions. From a previous study [24] it is known that the fatigue strength of Ti-8.6Al at 350°C for the coarse grained microstructure in vacuum is about 400 MPa which is even higher than the fatigue strength of the fine grained material in air (320 MPa). Obviously, the same improvement of the fatigue strength by shot peening and recrystallization as reported in the present work on rotating beam loaded specimens, could be achieved in push-pull loading (no stress gradient).

For the (α+β) Ti-6242 alloy the modification of the near surface microstructure from a fine lamellar to a fine equiaxed morphology of the α and β phases led to a similar improvement of the elevated temperature fatigue strength (Figure 6) as the grain refinement in Ti-8.6Al (compare Figure 6 with Figure 3). The increase in the 10^7 cycles fatigue strength at 550°C from about 390 to 450 MPa can be attributed to an improved resistance to the nucleation of fatigue cracks because of the higher yield stress of the equiaxed compared to the lamellar microstructure at 550°C (Table 2). From parallel work [25] in which the fatigue behavior of the Ti-6242 alloy at 550°C was studied on specimens having an equiaxed microstructure in the bulk it is known that the 10^7 cycles fatigue strength at 550°C is about 425 MPa and thus somewhat lower than the fatigue strength of the shot peened and recrystallized specimens (450 MPa). This result may indicate that the yield stress in the surface layer of these shot peened and recrystallized specimens is probably higher than the value listed for the equiaxed microstructure in Table 2. A difference in yield stress between the equiaxed surface layer microstructure of the shot peened and the bulk equiaxed microstructure (listed in Table 2) can result from the different deformation temperatures and deformation degrees used before the recrystallization treatment.

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