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Improvements in the fatigue strength of Ti–6Al–4V through microstructure control

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Fatigue deformation and stage I (shear mode) crack initiation in Ti–6Al–4V alloy test pieces have been studied using optical microscopy. Two types of stage I fatigue crack initiation were observed, (a) along α/β interfaces and (b) transcrystalline initiation across α grains in partly transformed microstructures and across α/β interfaces in fully transformed microstructures. The α/β interface cracking occurred predominantly in the low stress regions of the test pieces. These observations suggested that a microstructure with a small α grain size, to minimize the mean free slip path, and with minimum lengths of α/β interface, would have a high fatigue strength. Such a microstructure, with an α grain size of < 10 μm, and spheroidal or near spheroidal β particles, was produced by thermo-mechanical processing. The rotating cantilever fatigue strength of this microstructure, $\pm$ 670 MN m^{-2} at 10^7 cycles, compares with fatigue strengths in the range $\pm$ 480 to $\pm$ 590 MN m^{-2} for commercial Ti–6Al–4V bars.

1. Introduction
A recent investigation [1] of the effect of section size on the fatigue strength of annealed Ti-6Al–4V bars showed that the axial loading fatigue strength of this alloy could range from $\pm$ 350 to $\pm$ 590 MN m^{-2} at 10^7 cycles. Small α grained microstructures found in small diameter bars had the highest fatigue strengths. In contrast, thick section forged or rolled bars, containing large volumes of similarly oriented α, had fatigue strengths at the lower end of the range. It was concluded that the differences in fatigue strength were due primarily to differences in ease of stage I (shear mode) crack initiation and growth in different microstructures. This paper concerns some observations on stage I fatigue crack initiation in Ti–6Al–4V, and gives the results of work, based on these observations, to synthesize a high fatigue strength microstructure in this alloy.

2. Material
For the metallographic observations on stage I crack initiation and growth, test pieces were cut from 235 mm wide × 57 mm thick forged Ti–6Al–4V bar, and from 1.6 mm thick rolled Ti–6Al–4V sheet. Experiments to develop a high fatigue strength microstructure were carried out on a 30 mm diameter rolled Ti–6Al–4V bar. The compositions of these materials are given in Table I.

3. Metallographic observations on stage I crack initiation
Torsion fatigue test pieces were used for the metallographic observations on crack initiation. Extensive stage I crack initiation and growth is favoured in this type of test piece, as the tensile component across the faces of developing stage I cracks is small. Therefore, cracks continue to grow by a stage I growth mechanism, instead of rapidly changing to stage II (tensile mode) growth, as is the case for rotating cantilever, reversed plane bending, or direct stress test pieces.

Test pieces cut from the 57 mm thick forged and annealed (2 h at 700°C) bar were electropolished and torsion fatigued at room temperature. Examination of the electropolished surfaces revealed stage I crack initiation at α/β interfaces (indicated A, Fig. 1). There was also evidence of transcrystalline or transinterface deformation (B, Fig. 1) and cracking (A, Fig. 2). α/β interface cracking predominated in the low stress regions

TABLE I Analysis of Ti–6Al–4V materials used in the investigation (wt %)

Material	Al	V	Fe	C	H	O_2	N_2
235 mm wide × 57 mm thick forged bar	6.12	3.97	0.07	0.03	0.001	0.072	0.013
1.6 mm thick rolled sheet	6.00	3.97	0.11	0.005	0.0051	0.15	0.009
30 mm diameter rolled bar	5.89	3.99	0.19	0.019	0.0048	0.13	0.007

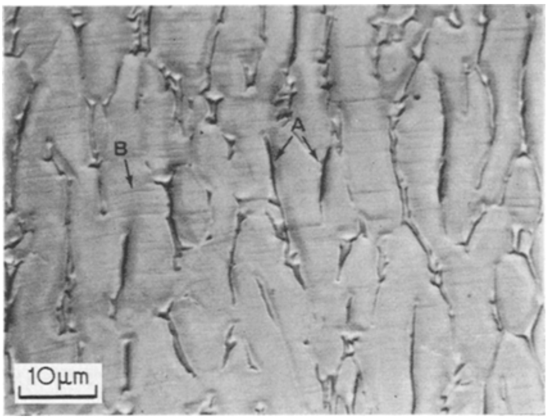

Figure 1 Crack initiation along α/β interfaces, A, and transcrystalline deformation, B, on an electropolished and torsion fatigued surface of the 57 mm thick Ti–6Al–4V bar.

of the torsion fatigue test pieces, while in the high stress regions, both interface and trans-interface cracking were observed. As the stress intensity factor range, ΔK (where $\Delta K = K_{max} - K_{min}$), of growing stage I interface cracks increases, growth probably changes to the

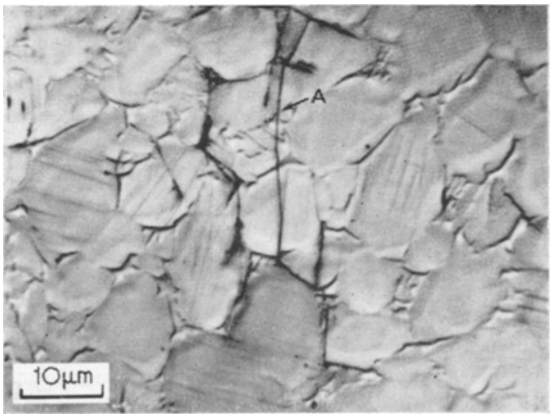

Figure 2 Transcrystalline cracking, A, on an electropolished and torsion fatigued surface of the 57 mm thick Ti–6Al–4V bar.

stage I transinterface mode. With further increases in ΔK values, transition to the stage II crack growth mode will occur, but not before a critical value, determined previously to be ~ 13 MN $m^{-3/2}$ [1], has been exceeded.

To study interface cracking further, torsion fatigue test piece blanks cut from 1.6 mm thick Ti–6Al–4V sheet were heat-treated just above the β transus. Furnace cooling then gave a coarse transformed β microstructure with numerous long α/β interfaces. These test pieces were also electropolished and fatigued at room temperature. Interface (A, Fig. 3) and trans-

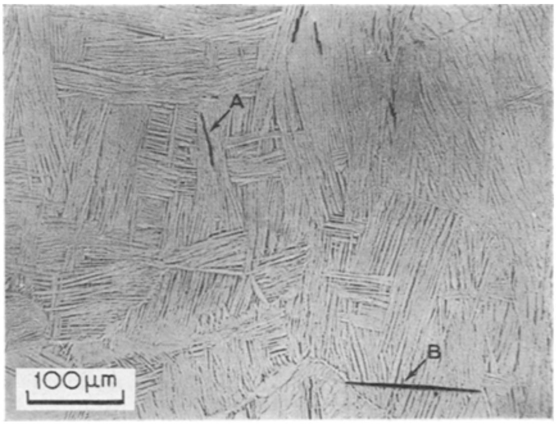

Figure 3 Crack initiation along α/β interfaces, A, and across α/β interfaces, B, on an electropolished and torsion fatigued surface of the 1.6 mm thick Ti–6Al–4V sheet.

interface (B, Fig. 3) crack initiation were again observed. Interfaces that were particularly favourable for stage I crack initiation were those along prior β or sub-grain boundaries (Fig. 4).

Damage bands were observed ahead of some transcrystalline or transinterface stage I crack tips. Such a band is indicated A in Fig. 5. It can be seen that the cutting of β plates results in the formation of localized preferred paths for further deformation. These damage bands appear to be analogous to those observed earlier

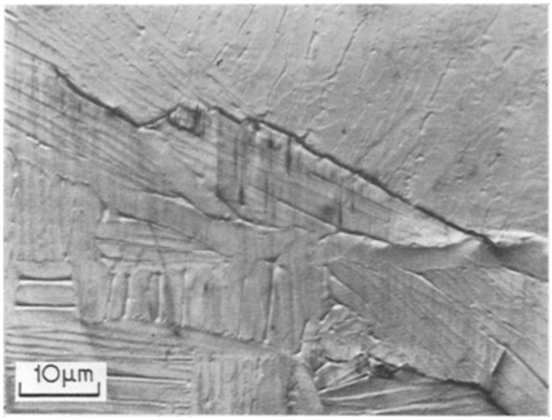

Figure 4 Interface cracking along a grain boundary on an electropolished and torsion fatigued surface of the 1.6 mm thick Ti–6Al–4V sheet.

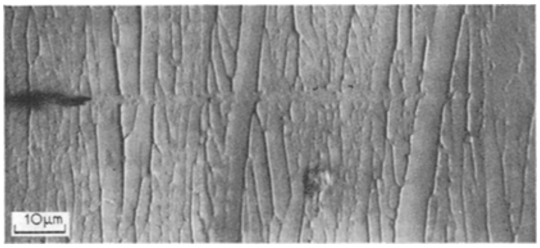

Figure 5 Damage band, A, ahead of a stage I crack tip on an electropolished and torsion fatigued surface of the 1.6 mm thick Ti–6Al–4V sheet. Specimen subsequently re-electropolished.

in aluminium–zinc–magnesium alloys [2], where reversed slip associated with stage I crack initiation resulted in precipitate cutting and re-solution.

These observations indicate that stage I initiation can occur with relative ease in Ti–6Al–4V when the microstructure contains long β plates with associated long α/β interfaces, and the grain size is large, giving a long mean free slip path for the development of transcrystalline damage bands. High fatigue strength would, therefore, be expected from a Ti–6Al–4V microstructure consisting of very small α grains, with the β phase in the form of spheroidal or near-spheroidal particles. Additionally, from crystallographic considerations, the texture should be such that the prism planes, on which transcrystalline initiation is most likely to occur, should be aligned so that the resolved shear stress on them is as small as possible. If the grain size is small, the amount of twinning, and hence possible cracking at twin interfaces, will be virtually eliminated. With these considerations in mind an attempt was made to synthesize a high fatigue strength microstructure in the Ti–6Al–4V alloy.

4. Production of a high fatigue strength microstructure

To obtain a uniform distribution of spheroidal or near-spheroidal β phase in the final bar it was considered desirable that the texture before working should be random, and that the microstructure should be a homogeneous one consisting of β plates in a matrix of secondary α. To obtain this microstructure a length of 30 mm diameter bar was β annealed at 1030° C, furnace cooled to 800° C and then air cooled. To obtain a small α grain size in the final bar it was considered necessary to work the bar at a temperature low in the $\alpha + \beta$ field. It was also anticipated that low temperature working would spheroidize the β phase and reduce its volume fraction in the microstructure, and also ensure that β transformation products were not formed. Thus the number, and particularly the length, of α/β interface crack initiation sites would be reduced to a minimum. Accordingly the 30 mm diameter bar was rolled to 9.4 mm square bar in the temperature range 700 to 650° C. The microstructure of the 9.4 mm square bar obtained by the sequence outlined is shown in Fig. 6. The small α grain size and the spheroidized β can be clearly seen. (0001) and $(10\bar{1}0)$ pole

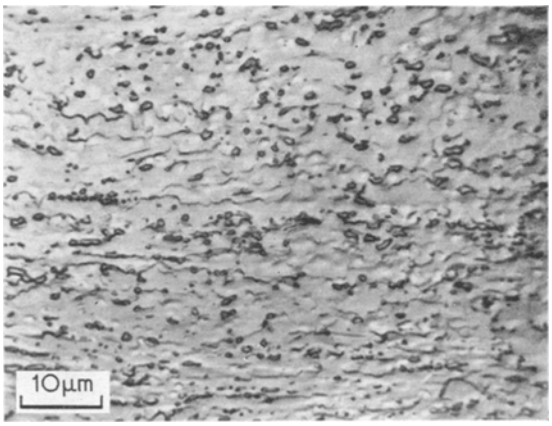

Figure 6 Microstructure of the 9.4 mm square Ti–6Al–4V bar produced by an 8:1 rolling reduction in the temperature range 700 to 650° C, and subsequently annealed at 700° C. Etched longitudinal section.

TABLE II Tensile properties of the 30 mm diameter and 9.4 mm square Ti–6Al–4V bars

Bar	Condition	0.2% proof stress ($MN\ m^{-2}$)	Tensile strength ($MN\ m^{-2}$)	E ($\times 10^{-3}$) ($MN\ m^{-2}$)	% Elongation on 5.65 $\sqrt{A}$
30 mm diameter	β annealed at 1030° C, furnace cooled to 800° C, then air cooled	838	948	115.3	12.5
9.4 mm square	$\alpha\beta$ worked in temperature range 700 to 650° C. Annealed at 700° C	1036	1147	110.3	12.5

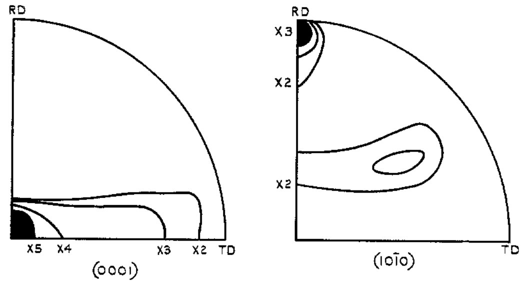

Figure 7 Pole figures of a longitudinal section of the 9.4 mm square Ti–6Al–4V bar produced by rolling in the temperature range 700 to 650° C, and subsequently annealed at 700° C.

figures were determined for this bar and are shown in Fig. 7.

5. Tensile tests

The tensile properties of the 30 mm diameter bar after β annealing, and of the 9.4 mm square $\alpha\beta$ worked bar annealed at 700° C were determined. 4 mm diameter, 33 mm parallel gauge length, $\frac{1}{4}$ in. BSF tensile test pieces were used and the results are given in Table II.

6. Fatigue tests

S–N curves were determined for the β annealed bar, and for the $\alpha\beta$ worked and annealed bar,

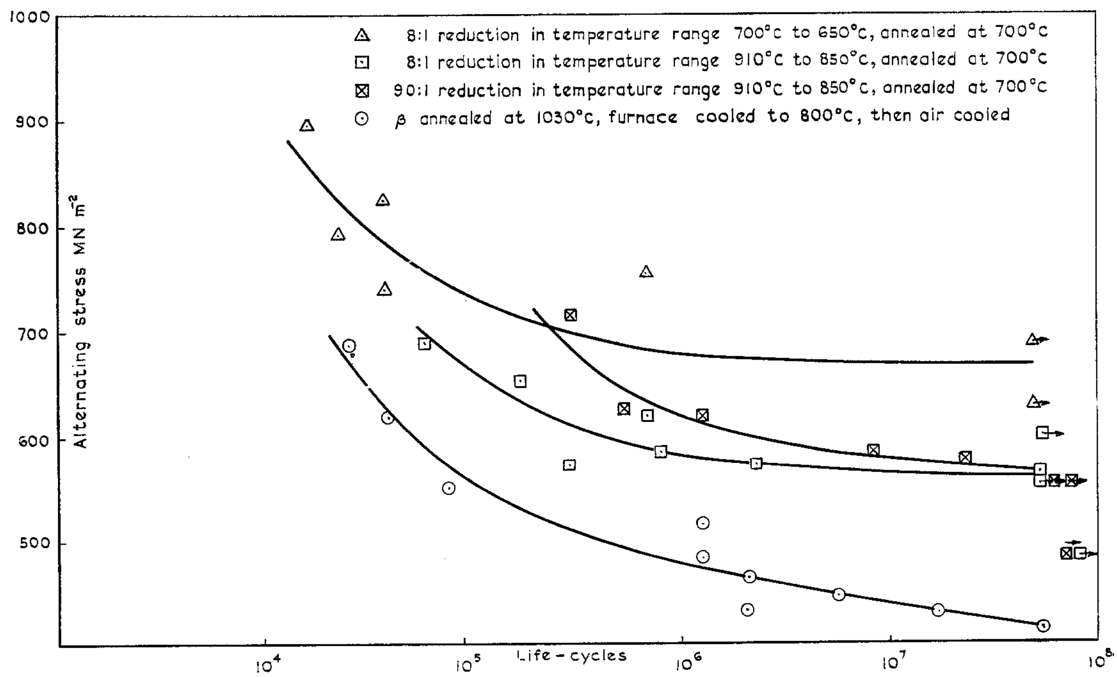

Figure 8 S–N curves for rotating cantilever fatigue tests on Ti–6Al–4V bars β annealed, and β annealed and $\alpha + \beta$ worked at high and low temperatures in the $\alpha + \beta$ field, and subsequently annealed.

using rotating cantilever Rolls Royce fatigue test pieces, 4 mm minimum diameter, 51 mm long. The results obtained are plotted in Fig. 8. It can be seen that the fatigue strength at 10^7 cycles of the β annealed material was $\pm$ 440 MN m^{-2}, and that of the material given the low temperature 8:1 $\alpha\beta$ reduction was $\pm$ 670 MN m^{-2}. Thus the fatigue ratio (defined as the fatigue strength at 10^7 cycles/tensile strength) of the two materials was, respectively, 0.46 and 0.58.

7. Discussion

Observations [3-5] on the Ti–6Al–4V alloy, including those reported here, have revealed two types of stage I fatigue crack initiation. Cracks developed either across α grains or in transformed structures across $\alpha\beta$ interfaces, or they initiate along the $\alpha\beta$ interfaces. In highly stressed regions both types of crack initiation occurred, while in regions of low stress, $\alpha\beta$ interface cracking predominated. This suggested that to obtain good high cycle fatigue strength, the number and particularly the length, of the $\alpha\beta$ interfaces available as crack initiation sites should be reduced to a minimum. In the present attempt to obtain a high fatigue strength microstructure this condition was fulfilled by low temperature working which resulted in a low volume fraction of spheroidal or near-spheroidal β. Low temperature working also gave a very small α grain size, thereby ensuring a small mean free slip path, a further requirement for high fatigue strength [6-8].

In an earlier investigation [8], an improvement in the low cycle fatigue strength of Ti–6Al–4V was associated with a refined α grain size and with the presence of over-aged martensite in the microstructure. The high cycle fatigue strength was, however, not significantly better than that of a microstructure with a larger α grain size and containing a coarse β transformation product. In this earlier investigation controlled $\alpha + \beta$ reductions were given to Ti–6Al–4V alloy bars by working in the temperature range 910 to 850°C. The reductions were 3, 8, 32 and 90:1. The S–N curves for the 8 and 90:1 reductions are included in Fig. 8, and the association microstructures are shown in Figs. 9 and 10. The refined α grain size and the over-aged martensite in the microstructure of the bar given the 90:1 reduction can be clearly seen in Fig. 10. The increased low cycle fatigue strength with very little improvement in high cycle strength can be seen in Fig. 8.

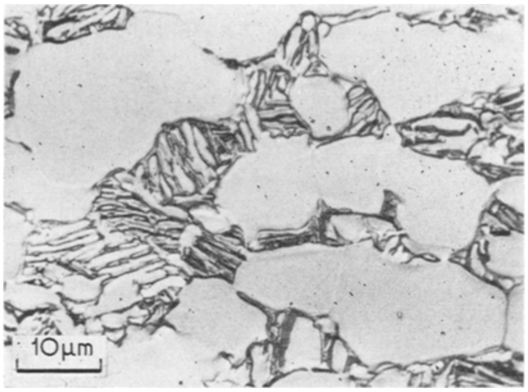

Figure 9 Microstructure of a 36 mm square Ti–6Al–4V bar produced by an 8:1 rolling reduction in the temperature range 910 to 850°C, and subsequently annealed. Etched longitudinal section.

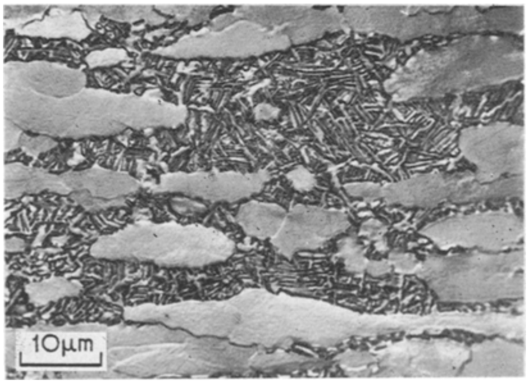

Figure 10 Microstructure of a 11 mm square Ti–6Al–4V bar produced by a 90:1 rolling reduction in the temperature range 910 to 850°C, and subsequently annealed. Etched longitudinal section.

Comparison of Fig. 9 with 6 shows the microstructural difference between the bars given an 8:1 reduction at high (910 to 850°C) and low (700 to 650°C) temperatures. With high temperature working there is a larger proportion of β phase in the microstructure at the working temperature. The transformation product formed from the β phase depends on the cooling rate from that temperature. Thus, the higher the final working temperature, the greater the proportion of transformation product, and the slower the cooling rate, the coarser this product will be. If the final working temperature is above the M_s ($\sim$ 830°C), and the cooling rate is sufficiently rapid, then the microstructure will be partly martensitic [9]. This was the case for the 90:1 high temperature reduction. The

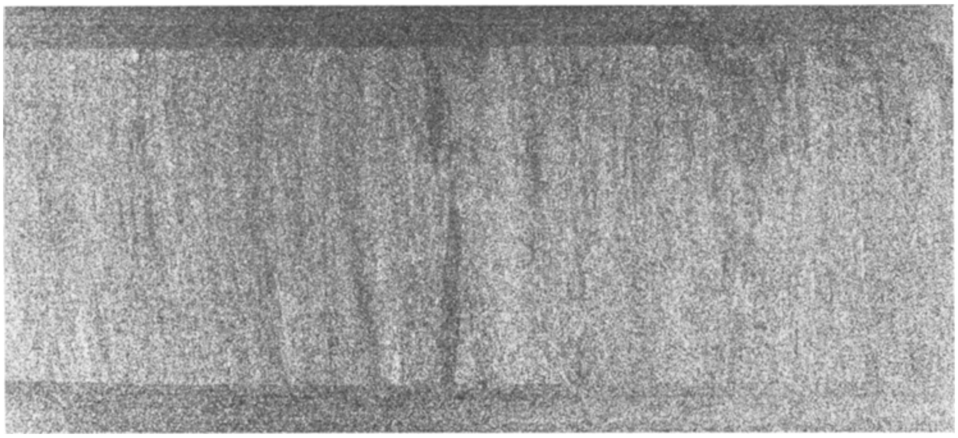

Figure 11 Macrostructure of slice from the 57 mm thick Ti–6Al–4V bar clad with Ti–6Al–4V sheet (× 5).

as-worked microstructures of this, and the other bars, were modified by subsequent annealing at 700° C, but the extensive linear $\alpha\beta$ interfaces of the decomposed martensite can be clearly seen in Fig. 10. These are in marked contrast to the limited extent of the $\alpha\beta$ interfaces in the bar worked at low temperature, due to the spheroidal or near-spheroidal form of the limited amount of the β present in this bar. The α grain size of the 90:1 high temperature reduction was, however, very similar to that obtained from the 8:1 low temperature reduction. The improvement in the high cycle fatigue strength of the latter material is therefore attributed to the reduction in number and length of $\alpha\beta$ interfaces available as stage I crack initiation sites.

In contrast to the marked effect of working temperature on the microstructure of the Ti–6Al–4V bars, very little difference could be detected between their textures, or between the textures of these bars and that of a commercially produced 30 mm diameter bar [10]. The present improvement in fatigue strength has, therefore, been achieved largely by microstructural modification. The possibility therefore exists of a further increase in fatigue strength if a texture can be obtained to reduce the proportion of prism planes lying close to the directions of maximum resolved shear stress, without increasing the possibility of twin interface initiation.

Very small α grains and spheroidal β will be very difficult to obtain throughout large forgings, but as fatigue cracks are predominantly surface initiated, another approach to improving the fatigue strength of large forgings would be to modify their surface microstructure to one of high fatigue strength. To achieve this, Ti–6Al–4V sheet with the required microstructure could be explosively bonded onto large forged components, and the interface improved by diffusion bonding. Shot peening such a layer would further improve its fatigue strength. Thick plates can be clad with high fatigue strength material by roll (Fig. 11), or diffusion bonding, but this must be carried out at a relatively low temperature to preserve the correct microstructure in the clad material. In such a process the core material would first be heat-treated to optimize toughness, and such a duplex structure would then have a combination of high fatigue strength and high toughness. Alternatively, thick plate can be built up by diffusion bonding together sheet with the necessary high fatigue strength microstructure.

8. Conclusions

1. In fully transformed Ti–6Al–4V microstructures stage I fatigue crack initiation can occur either across or along $\alpha\beta$ interfaces. $\alpha\beta$ interface initiation occurs predominantly at low stress, and the interfaces that are particularly favoured as crack initiation sites are those at prior β grain or sub-grain boundaries. When the microstructure contains some primary α, cracking can initiate across the primary α, or at $\alpha\beta$ interfaces. Again $\alpha\beta$ interface cracking is favoured at low stress.

2. Long mean free slip paths can be formed in totally transformed microstructures by β phase cutting in the zones of plastic deformation ahead of stage I fatigue cracks.

3. A Ti–6Al–4V microstructure produced by annealing at 700°C after an 8:1 reduction in the temperature range 700 to 650°C, and consisting of small α grains and spheroidal or near-spheroidal β, had a fatigue strength at 10^7 cycles of $\pm$ 670 MN m^{-2}. The tensile strength was 1147 MN m^{-2}, and thus the fatigue ratio was 0.58.

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