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THE EFFECT OF TESTING DIRECTION ON THE FATIGUE AND TENSILE PROPERTIES OF A Ti-6Al-4V BAR

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Introduction

This paper reports the effect of testing direction on the fatigue and tensile properties of a Ti-6Al-4V forged bar. Directionality of mechanical properties can be of considerable significance in worked h.c.p. materials, firstly, due to their crystal structure, which is inherently anisotropic and secondly, due to any grain elongation introduced by the working operation, where there can be marked differences in the number of grains per unit area in different directions but, if little or no recrystallization occurs, there can be no change in the number of grains per unit volume (Fig 1). Any effects due to the mechanical fibering of inclusions etc., which markedly affects the mechanical properties of most cubic materials (eg refs 1-3), is absent in titanium alloys because of the purity of the starting material, their method of preparation (vacuum arc melting) and since titanium dissolves most of the common alloying or inclusion forming elements.

Material Composition and Processing History

Al	V	Fe	C	O_2	N_2	H_2	wt%
6.12	3.97	0.07	0.03	0.072	0.013	0.001	

21 in diameter ingot. $\xrightarrow{\beta \text{ forged}}$ 16in square bar $\xrightarrow{\alpha+\beta \text{ forged } (\sim 950^{o}C)}$ $9\frac{1}{4}$in wide x $2\frac{1}{4}$in thick bar (final dimensions) bar then annealed at $700^{o}C$ for $2\frac{1}{2}$h.

Testing

Test pieces were machined from the annealed bar in the longitudinal, transverse and short transverse directions. The longitudinal direction was parallel to the direction of maximum extension during forging (Fig 1). S-N curves were determined on longitudinally polished Rolls Royce rotating cantilever test pieces (2in long; 0.16in minimum dia) fatigued at a frequency of 6000 cycles/min. Tensile properties were determined on standard test pieces (1.3in gauge length; 0.16in dia) at a strain rate of 3×10^{-3} sec^{-1}.

Results

S-N curves are shown in Fig 2. The most noticeable features of this figure are the marked degree of scatter and the large variation in the fatigue strength with testing direction. Approximate fatigue strengths at 10^7 cycles are given in Table I.

Tensile properties for the three directions are also listed in Table I. Yield and tensile strengths in both the transverse and short transverse directions were considerably greater than in the longitudinal direction. A modulus change was observed only in the transverse direction. Ductility values remained almost constant except for a small decrease in the elongation value for the short transverse direction.

Table I - Mechanical Properties of the $9\frac{1}{4}$in wide x $2\frac{1}{4}$in thick forged and annealed Ti-6Al-4V Bar

Testing Direction	0.2% Proof Stress (ksi)	Tensile Strength (ksi)	E ($\times 10^{-3}$) (ksi)	Elong %	R of A %	Approx Fatigue Strength at 10^7 cycles (ksi)
Longitudinal	121.0	132.0	16.5	17.5	33	± 72
Transverse	135.5	143.0	18.7	17.0	30	± 62
Short transverse	129.5	142.0	16.5	12.5	30	± 82

Metallographic Observations

Microstructure

The grain elongation due to working could be clearly observed by examination of the etched bar at low magnifications (Fig 1). Although the bar showed small α grains surrounded by β at many of the grain boundaries, polarized light microscopy revealed clear evidence of a large prior β grain size, (ie large volumes of similar orientation) with each prior β grain containing many α grains (Fig 3). Only partial recrystallization had occurred, eg AA in Fig 3.

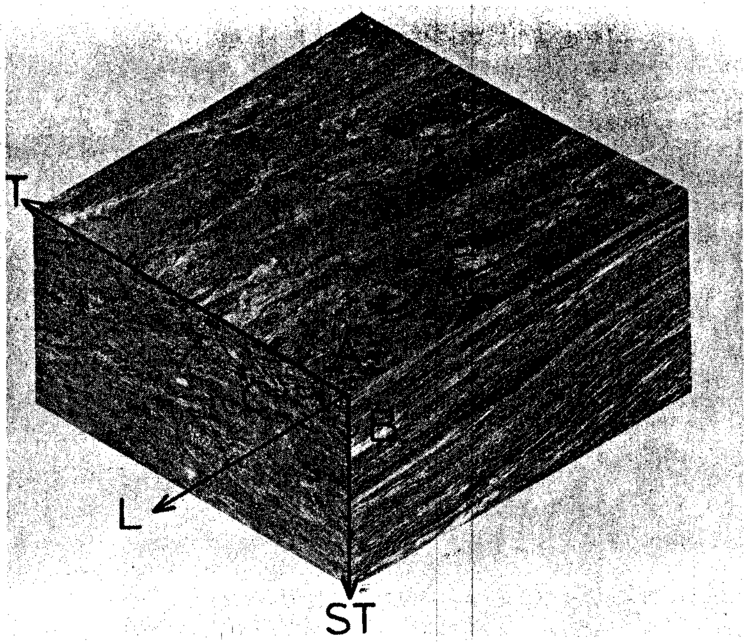

Fig. 1. Macrostructure in the three orthogonal sections of the 2¼ in thick forged and annealed Ti-6Al-4V bar showing the different number of grains per unit area in each plane. Etching solution 10% HF, 25% HNO_3, 65% H_2O. L, T, and ST - longitudinal, transverse and short transverse directions respectively. X2

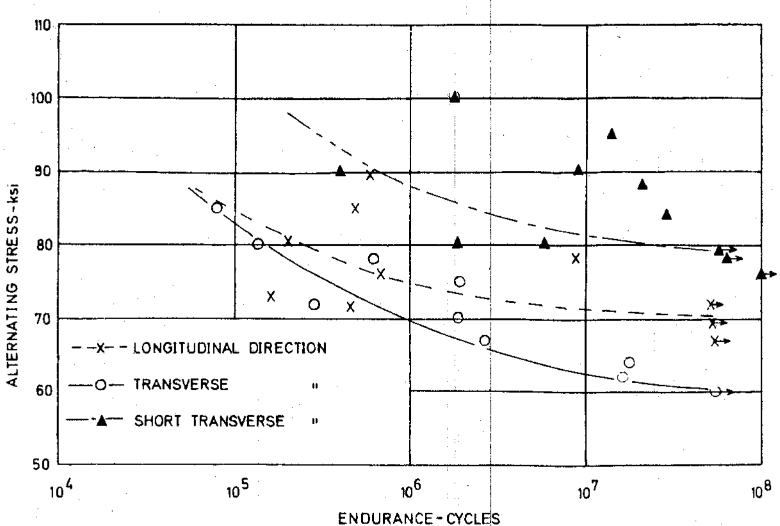

Fig. 2. The effect of testing direction on the fatigue life of the 2¼ in thick forged and annealed Ti-6Al-4V bar.

Many of the elongated regions observed on face B, Fig 1 were found to be up to 80 VPN harder than more equiaxed regions. These hardness variations can be attributed to crystallographic and/or chemical differences. Microprobe analyses revealed higher vanadium contents in the harder zones (by ~ 0.6 wt %) and also smaller aluminium variations (~0.2 wt %). However, in view of the texture analyses (see following section) which showed that a large proportion of basal plane poles are normal to face B, Fig 1, and hence would lead to hardness variations of the order of magnitude observed (4), it is most probable that crystallographic differences are primarily responsible for the hardness variations, with compositional differences enhancing this effect.

Texture

Considerable variations in preferred orientation were observed along and through the thickness of the bar and no one pole figure could be claimed to be representative of the bar as a whole. From the pole figures determined, however, it was apparent that at the surface of the bar (ie face A, Fig 1) most of the basal planes were approximately parallel to the surface; below the surface and through most of the thickness of the bar basal planes were mostly parallel to the bar edge (ie face B, Fig 1). In all cases the $<10\bar{1}0>$ direction was in the longitudinal direction. Between these two textures it was possible to record a wide range of preferred orientations. The variations in texture are considered to be due to inhomogeneous deformation caused by temperature gradients or inhomogeneous working introduced into the bar during forging.

Fatigue Crack Initiation and Propagation

Fatigue cracks formed in α grains, usually in equiaxed regions such as AA in Fig 3. The extent of stage I or shear mode growth (5) seemed to be dictated by the applied stress amplitude for test pieces from all three directions. This is in agreement with other investigations (6,7).

Local variations in stage II growth rates were readily apparent on examination of fracture surfaces by optical and scanning electron microscopy. This was most noticeable on longitudinal and transverse test piece fracture surfaces, particularly when crack growth was in the short transverse direction. In these cases locally straight crack fronts could be observed frequently as faster growth in some prior β grains could be seen to terminate at prior β grain boundaries as the crack front entered a less favourably oriented prior β grain (Fig 4(a)). For growth in the longitudinal and transverse directions, serrated crack fronts were often observed (Fig 4(b)). Here faster growth in a favourably oriented prior β grain is held back by slower growth in less favourably oriented grains on either side.

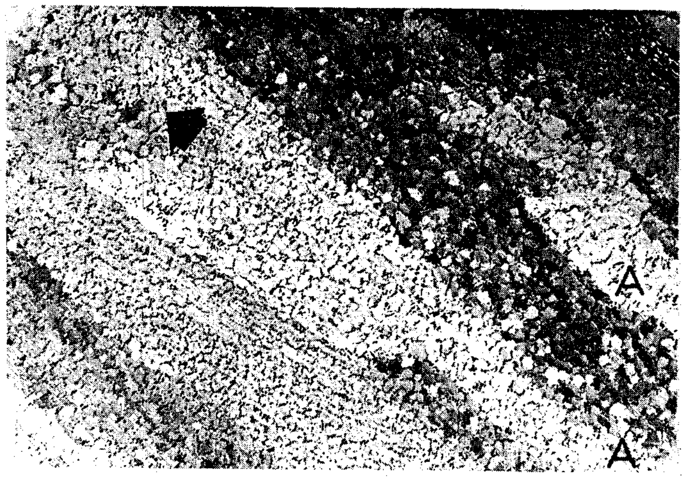

Fig. 3. Polarized light micrograph showing marked prior β grain size in the $2\frac{1}{4}$ in thick forged and annealed Ti-6Al-4V bar. Electropolished longitudinal section, longitudinal direction shown by arrow. X120

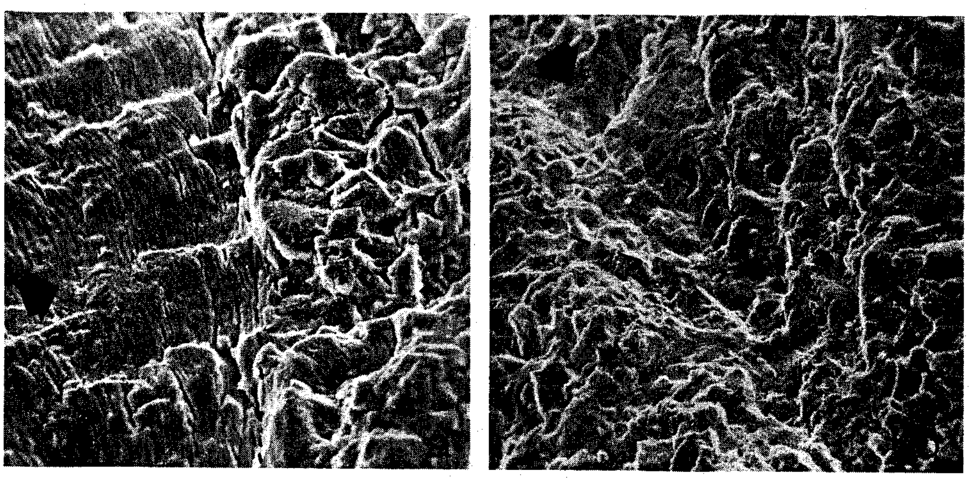

Fig. 4. Scanning electron micrographs showing fatigue crack growth in a longitudinal fatigue test piece cut from the $2\frac{1}{4}$ in thick forged and annealed Ti-6Al-4V bar (a) in the short transverse direction (X700) and (b) in the transverse direction (X340). Crack growth directions shown by arrows. Crack fronts shown by dotted lines.

Different growth rates, with changes in orientation, are not unexpected since growth is dependent, among other factors, on:-

(i) the plastic zone size, which is inversely related to yield strength (8), and changes in yield strength with orientation are common in h.c.p. materials (4, 9),

and (ii) Elastic modulus, which enters into many crack growth equations (eg 10, 11), usually as a measure of strain (ie $\frac{\text{stress term}}{E}$) and E has been shown to be very sensitive to crystal orientation (4, 12).

Crack propagation measurements on 0.5 in thick edge notched test pieces cut from $2\frac{1}{4}$in thick rolled and annealed Ti-6Al-4V bar have shown little difference in the macroscopic rate of stage II growth with testing direction, for ΔK values in excess of $\sim$ 14 ksi $\sqrt{\text{in}}$ (13). Since stage II growth rates in forged and rolled bars would not be expected to be very different, it can be assumed that the overall stage II growth rate in the forged bar used in this investigation is also independent of testing direction, and that the macroscopic growth rates average out the local variations described above.

Texture analysis of individual test pieces, in addition to the general studies described earlier, reaffirmed the texture variability, and revealed that test pieces oriented with a large proportion of prism planes approx parallel to the stress axis had shorter fatigue lives than those with a large spread of orientations or with a large proportion of basal planes approx parallel to the stress axis.

It may therefore be concluded that the differences in fatigue strength with testing direction arise predominantly from differences in the ease of crack initiation and stage I growth.

Tensile Deformation

Anisotropic deformation behaviour was common to tensile test pieces from all three directions in the bar. The effect of the prior β grain size could be observed to dictate the extent of slip (Fig 5). The effect of preferred orientation was apparent in the non-uniform necking of all test pieces. Minimum to maximum diameters of the neck were found to be in the range 0.75-0.85. This is due to the large proportion of basal planes lying approx parallel to the stress axis, slip occurs on prism planes and little contraction takes place in the c-direction. Examples of such behaviour are shown in Fig 6. Slip on intersecting prism planes, at 60° to one another, could be observed on many of the fracture surfaces.

There was evidence of a more uniform reduction of area in transverse test pieces where many of the basal planes were perpendicular to the stress axis. (C, Fig 6.) A modulus of 18×10^6 psi was recorded in this case (Table I). However, even then there must still be a considerable number of basal planes at angles $< 90^\circ$ to the stress axis, since a value of $E = 21 \times 10^6$ psi would be expected if all basal planes were normal to the stress axis (4, 12). In these transverse test pieces, extensive $\{10\bar{1}2\}$ twinning occurred in favourably oriented prior β grains which, again, was limited by the size of those grains (Fig 7). It should be noted that it is elongated grains, such as that shown in Fig 7, which give rise to the hardness variations described earlier.

Discussion

This investigation has shown that considerable variations in fatigue and tensile strength can occur with changes of testing direction in large forgings of Ti-6Al-4V. In the following discussion explanations will be proposed to account qualitatively for this mechanical anisotropy in terms of texture and grain shape.

Fatigue Behaviour

Consideration of texture and stressing mode might lead to the conclusion that there should be no effect of texture during the early stages of fatigue, since a tension-compression cycle parallel to the basal plane is equivalent to a compression-tension cycle normal to the basal plane. However, this is not strictly the case since the elastic and plastic Poisson's ratios (μ_e, μ_p) are very sensitive to crystal orientation (12, 14). Thus a strain applied parallel to the basal plane will, on average, result in smaller strain changes normal to the stress axis (because of the higher values of μ_e and μ_p) than an equivalent strain applied perpendicular to the basal plane. It may therefore be proposed that on fatiguing test pieces with basal planes parallel to the stress axis (longitudinal and short transverse directions), the constraining influence of the higher Poisson's ratios enhances the fatigue life compared with test pieces with basal planes normal to the stress axis (transverse direction). Experimental results and texture analyses are in agreement with this suggestion.

It then remains to be explained why the longitudinal direction, while nominally having the same texture as the short transverse direction, has a lower fatigue strength. This is to be expected in view of its lower tensile strength (Table I) and is considered to be due to the changes in grain shape caused by working (Fig 1) and their orientation with respect to the stress axis. Longitudinal test pieces have the large volumes of similar orientation oriented along the length of test pieces and hence there is a greater probability that a smaller

Fig. 5. Polarized light micrograph showing prior β grain size limiting extent of surface slip. Electropolished short transverse tensile test piece cut from the 2¼ in thick forged and annealed Ti-6Al-4V bar. Stress axis shown by arrow. X330

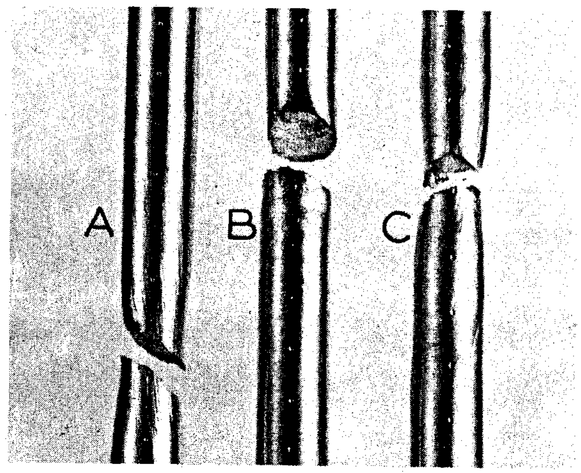

Fig. 6. Photograph showing deformation on longitudinal (A), short transverse (B) and transverse (C) tensile test pieces cut from the 2¼ in thick forged and annealed Ti-6Al-4V bar. X2

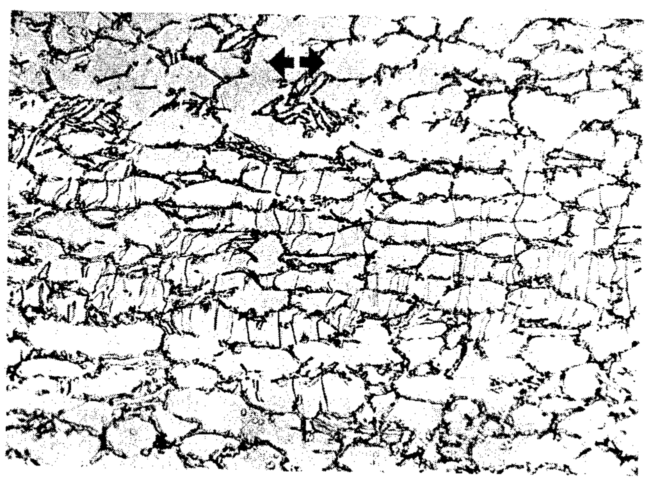

Fig. 7. Photomicrograph showing prior grain size limiting extent of $\{10\bar{1}2\}$ twinning. Electropolished and etched longitudinal section of a transverse tensile test piece cut from the 2¼ in thick forged and annealed Ti-6Al-4V bar. Stress axis shown by arrow. Etching solution 1% HF, 12% HNO_3, 87% H_2O. X500

number of these prior β grains per unit area are at the surface thus leading to easier crack initiation and stage I growth and resulting in a lower fatigue strength. This effect of grain shape is also probably a contributory factor to the low transverse fatigue strength, complementing the effect of texture.

The considerable scatter in Fig 2 can be attributed to the variable texture and coarse macrostructure common to all three directions. Variable surface finish is not the cause of scatter, firstly, since further working (7) and heat treatment (15) of this bar, using the same test piece preparation techniques and surface finish, produced very much less scatter, and secondly, electropolished test pieces cut from the as received bar showed approximately the same degree of scatter as mechanically polished ones. Small local variations in oxygen content are also a possible source of scatter.

Tensile Behaviour

Anisotropy in tensile properties was only noticeable in changes in strength (Table I). Little influence of crystallography on ductility was observed since, even when extensive twinning was favoured, as in the transverse direction, it can be shown that slip accounted for nearly all the deformation (9).

The effect of the bar textures can be correlated closely with the changes in yield and tensile strength values. In transverse test pieces (basal planes normal to the stress axis), a higher modulus and yield strength and a smaller increase in tensile strength, compared with the longitudinal and short transverse directions, would be expected from theoretical considerations (4, 9, 12, 14). Experimental results are in good agreement with such predictions (Table I). The lower tensile strength in the longitudinal direction is believed to reflect the smaller number of prior β grains per unit area in these test pieces compared with test pieces cut in the short transverse direction, which has the same nominal texture as the longitudinal direction. The large volumes of similar orientation extended over considerable distances along the gauge length in the longitudinal direction, with the result that lower strengths were observed in this direction.

Conclusions

1 The rotating cantilever fatigue strength at 10^7 cycles of test pieces cut in the longitudinal, transverse and short transverse directions of a $9\frac{1}{4}$in wide x $2\frac{1}{4}$in thick forged and annealed Ti-6Al-4V bar are approx ±72, 62 and 82 ksi respectively. Explanations are proposed to account for these differences in terms of the predominant texture in each direction and also the change in grain shape (ie grain elongation) due to the

working operation, which results in different numbers of prior β grains per unit area in the three testing directions.

2 Considerable scatter in the fatigue lives were observed at all stress levels for all three tested directions. The scatter is attributed to the variable texture and coarse macrostructure of the bar.

3 Yield strengths for the longitudinal, transverse and short transverse directions of this bar were 121.0, 135.0 and 129.5 ksi respectively, and the tensile strengths for these directions were 132.0, 144.0 and 142.0 ksi respectively. These trends are also explained in terms of the predominant texture in each direction and the different number of prior β grains per unit area in the three testing directions.

4 The modulus increased from 16.5×10^6 psi for the longitudinal and short transverse directions to 18.7×10^6 psi for the transverse direction. No large changes in ductility were observed with testing direction. These observations can be explained in terms of the predominant texture for each testing direction.

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References

1 Zinkham, R E, "Longitudinal and Short Transverse Fatigue and Fracture Properties of Heavy Aluminium Alloy Plates, Produced by Forging and Rolling." Trans. Met Soc AIME Vol 245 July 1969 pp 1519-1525.

2 Gazza, G E and Larson, F R, "Impact Properties of Air and Vacuum Melted AISI 4340 Steel" ASM Trans Quart Vol 58 June 1965 pp 183-194.

3 Lauchener, E A "The Microstructure and Ductility of 17-4PH and 15-5PH Steels" J of Materials Vol 5 March 1970 pp 129-139.

4 Roberts, W T "Preferred Orientation and Anisotropy in Titanium" J Less Comm Metals Vol 4 Aug 1962 pp 345-361.

5 Forsyth, P J E "A Two Stage Fatigue Fracture Mechanism" Proc Crack Propagation Symposium Cranfield 1962 Vol 1 pp 76–94

6 Stubbington, C A and Bowen, A W "The Effect of Section Size on the Fatigue Properties of Ti–6Al–4V Bars" to be presented at the Second International Conference on Titanium, Boston, May, 1972.

7 Bowen A W and Stubbington C A "The Effect of $\alpha + \beta$ Working on the Fatigue and Tensile Properties of Ti–6Al–4V Bars" to be presented at the Second International Conference on Titanium, Boston, May, 1972.

8 Rice J R "Mechanics of Crack Tip Deformation and Extension by Fatigue" American Society for Testing and Materials STP 415 (1967) pp 247–309.

9 Reed–Hill R E "Role of Deformation Twinning in the Plastic Deformation of a Polycrystalline Anisotropic Metal" Deformation Twinning Met Soc AIME Conf Proc. Vol 25 (eds R E Reed–Hill, J P Hirth and H C Rogers) Gordon and Breach N.Y (1964) pp 295–320.

10 McEvily A J and Johnston T L "The Role of Cross–Slip in Brittle Fracture and Fatigue" Proc First Int Conf on Fracture Sendai Japan Vol 2 1965 pp 515–546.

11 Pearson S "Fatigue Crack Propagation in Metals" Nature Vol 211 (5053) Sept 3 1966 pp 1077–1078.

12 Zarkades A and Larson F R "Elasticity of Titanium Sheet Alloys" The Science, Technology and Application of Titanium (eds R I Jaffee and N E Promisel) Pergamon Oxford 1970 pp 933–941.

13 Bowen A W, to be published.

14 Larson F R "Anisotropy of Titanium Sheet in Uniaxial Tension" ASM Trans Vol 57 1964 620–631.

15 Bowen A W and Stubbington C A "The Effect of Heat Treatment on the Fatigue and Tensile Properties of a Ti–6Al–4V Bar" to be published.