An Analytic Approach to Low Stress Fatigue Crack Growth in Titanium

J. L. Robinson, P. E. Irving, and C. J. Beevers
Department of Physical Metallurgy and Science of Materials,
University of Birmingham, Birmingham 15, U. K.

Introduction

At moderate fatigue stress levels variables such as yield strength, grain size, metallurgical structure, and load ratio are observed to have only small effects on the growth characteristics of a given material (1). However, at lower stresses these variables can have significant effects. This is illustrated in this paper by considering results obtained from tests on two titanium alloys having widely differing structures and properties.

Experimental

The two titanium alloys studied were supplied by I.M.I. Ltd., and designated Ti 115 and Ti 318. Ti 115 is commercially pure single phase α-titanium, and Ti 318 is an $\alpha+\beta$ alloy. Details of the different conditions in which the alloys were tested are shown in Table 1. Pin loaded single edge-notch specimens, having dimensions conforming to the requirements of the K calibration in Ref. (2), were used. Fatigue tests were performed using an Amsler Vibrophore, in room temperature laboratory air, at frequencies in the range 100-150 Hz. Fatigue crack lengths were monitored continuously by the potential drop method. Tension-tension loading, with R ratios (Lmin/Lmax) ranging from 0.07 to 0.7, was used.

Results and Discussion

Figure 1 shows the fatigue crack growth data obtained from two Ti 318 specimens and a Ti 115 specimen, all tested at R = 0.33. In Figure 2 the effects of grain size and R ratio on the growth

characteristics of Ti 115 are presented. Whilst the data in Figure 1 indicates similar overall growth tendencies for Ti 318 and Ti 115 there are distinct and reproducible variations with structure and ΔK levels.

Table 1

Alloy	Principal alloying elements	Heat treatment	Structure	Mean linear intercept grain size mm	0.2% Proof strength MN m^{-2}
Ti 115	0.10 wt.% oxygen	Annealed 3 hours 923K, furnace cool	single phase α	0.04	260
		Annealed 24 hours 1123K, furnace cool	single phase α	0.25	220
Ti 318	6 wt.% Aluminium 4 wt.% Vanadium	Annealed $\frac{1}{2}$ hour 1203K, air cool	α + 25 vol.% transformed β	α 0.0034 β 0.0022	835
		Annealed $\frac{1}{2}$ hour 1203K, water quench, Aged 8 hours 773K.	α + 50 vol.% tempered martensite	α 0.0035 tempered martensite regions 0.0021	950

Below a ΔK of ~ 10 MN m$^{-3/2}$ the Ti 318 α + tempered martensite structure exhibits a higher growth rate than the α + transformed β by a factor of up to four. In the same ΔK region (4 - 10 MN m$^{-3/2}$) all the fatigue fracture surfaces displayed deviations from the macroscopic crack plane on the scale of the microstructure. Figure 3a shows an example of relatively planar regions of the order of the grain size occurring over much of the fracture surface in a Ti 115 specimen, for a ΔK of ~ 6.4 MN m$^{-3/2}$. This effect has been reported previously (3,4), and termed grain orientation control (GOC) growth. Similar effects are observed in the Ti 318 α + transformed β structure in this ΔK range, Figure 3c, but were much less marked in the Ti 318 α + tempered martensite structure. At ΔK levels greater than ~ 10 MN m$^{-3/2}$ the fracture surfaces (Figure 3b, d) exhibited a more

amorphous appearance with the disappearance of the planar regions shown in Figure 3a, c. The fractographic evidence of Figure 3 indicates a marked influence of structure on fatigue crack paths at these low growth rates. It has been suggested (3,4) that GOC occurs when the scale of the plasticity at the crack tip is of the order of, or less than, the grain size. In the Ti 318 α + transformed β structure the transition from GOC to a more amorphous structure was fairly clearly defined and occurred when the reversed plastic zone size, r_p, $\left[r_p = \Delta K^2/12\pi\sigma_y^2 \ (5) \right]$ became approximately equal to the mean linear intercept α grain size. ($\Delta K = 8.5$ to 10 MN m$^{-3/2}$, $r_p = 0.0027$ to 0.0038 mm). The transition was more gradual in the Ti 115, however for $\Delta K = 10$ MN m$^{-3/2}$ the estimated value of r_p was 0.039 mm, close to the grain size of the Ti 115.

A possible explanation for the higher growth rate of the martensite containing Ti 318 specimens in the region where GOC was observed to be strongest in the α + transformed β structure, can be based on the following observations. Firstly, that the crack path was preferentially contained in the martensite phase, and secondly that the reversed plastic zone size was always very much larger than the structural elements of the tempered martensite. Above ~ 10 MN m$^{-3/2}$ the reverse plastic zone size exceeded the size of the α grains and tempered martensite regions. Here the growth rates in both Ti 318 structures were almost identical.

The results in Figure 2 show marked effects of grain size and R ratio on growth rates in a situation where the reverse plastic zone size was smaller than the grain size for ΔK's of less than ~ 10 MN m$^{-3/2}$ for curves 1 and 2, and always smaller for the coarse grained specimens, curve 3. The mode of crack growth in these specimens was

typified by Figure 3a, and involved both the formation of the relatively planar regions, and their interconnection. These results suggest that the planar regions are primarily influenced by ΔK, whilst their interconnection is more likely to be controlled by K_{MAX}. These conclusions are indicated by the lower growth rates that result in coarse grained material, or low R ratio tests, interconnection being made more difficult by larger grains or lower values of K_{MAX}.

In the light of these observations crack growth at low stresses cannot be solely interpreted in terms of ΔK. Microstructural influences are appreciable, and the present evidence suggests that interactions between crack tip plasticity and the microstructure are important factors. In particular the scale of crack tip plasticity with respect to structural element size is a significant variable influencing propagation behaviour.

Acknowledgement

References

1. Johnson H.H., and Paris P.C., Engng. Fracture Mech., 1, 1968, 3.

2. Brown W.F., and Srawley J.E., ASTM STP 410, 1966.

3. Birkbeck G., Inckle A.E., Waldron G.W.J., J. Mat. Sci., 6, 1971, 319.

4. Robinson J.L., and Beevers C.J., Presented at the Second International Conference on Titanium, Cambridge, Massachusets, 1972.
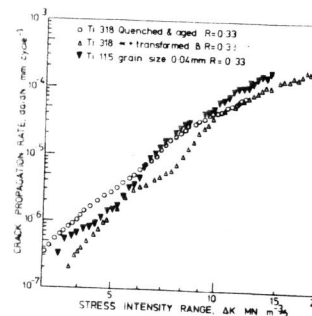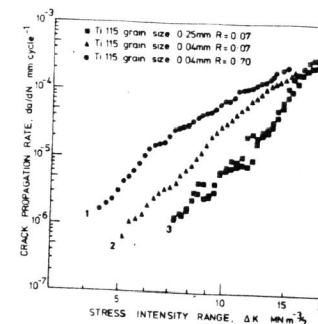
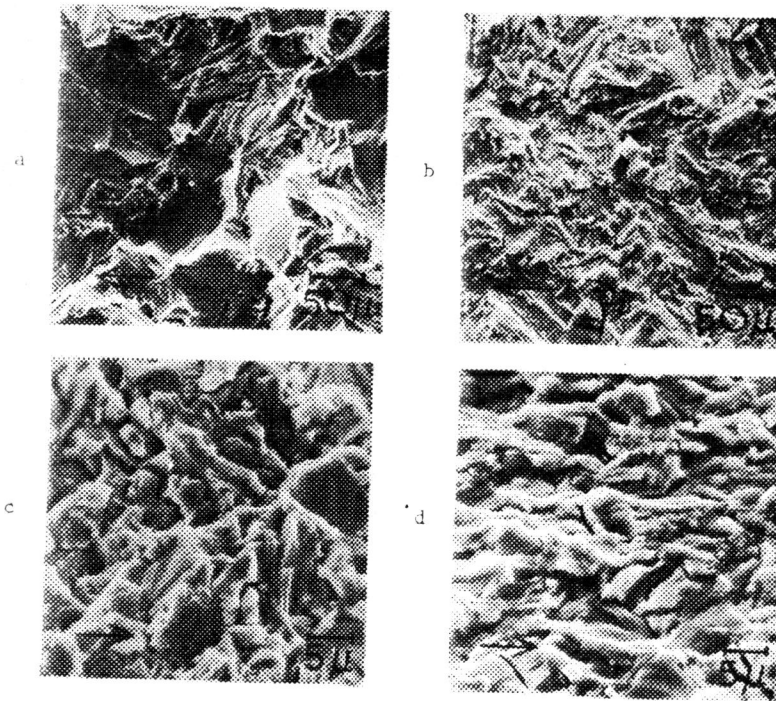5. Rice J.R., ASTM STP 415, 1967, 247.

Figure 1

Figure 2

Figure 3a. Ti 115, grain size ~ 0.04 mm, $\Delta K \sim 6.4$ MN m$^{-3/2}$
b. Ti 115, grain size ~ 0.04 mm, $\Delta K \sim 13.4$ MN m$^{-3/2}$
c. Ti 318, α + transformed β, $\Delta K \sim 5.4$ MN m$^{-3/2}$
d. Ti 318, α + transformed β, $\Delta K \sim 13.9$ MN m$^{-3/2}$