FRACTURE TOUGHNESS OF AN ISOTHERMALLY FORGED γ-TITANIUM ALUMINIDE

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The room temperature fracture toughness of as-forged, and forged and heat treated gamma-TiAl has been determined. The as-forged equiaxed-gamma microstructure gave lower toughness values. Heat treating to produce a fully lamellar microstructure increased the fracture toughness, but varying the colony size appeared to make little difference to the toughness. The K_{max} toughness values of these fully lamellar conditions produced by heat treatment after forging are similar to those obtained for as-cast fully lamellar microstructures produced by ingot casting alone.

INTRODUCTION

Gamma based titanium aluminides have a low density and good specific high temperature mechanical properties. They currently have a service temperature limit of about 700°C due to problems of oxidation and creep, and suffer from relatively low ductility, fracture toughness and steep fatigue crack growth resistance curves at ambient temperatures [1]. This work will examine the effect of isothermal forging on fracture toughness. The material studied here is Ti-48Al-2Mn-2Nb (in atomic per cent) which has been cast and subsequently forged.

EXPERIMENTAL

A section of height 133 mm was cut from a cast ingot of 90 mm diameter produced by plasma arc cold hearth melting at the University of Birmingham. The composition of the ingot (as determined by X-ray fluorescence at IMI Titanium) was Ti-48.1Al-1.9Mn-2.0Nb. Prior to forging the section was heat treated at 1250°C for 24 hours in a vacuum furnace to promote equiaxed-gamma formation. The material was isothermally forged in two stages at a temperature of 1150°C, a strain rate of 5×10^{-3} s^{-1}, to a total height reduction of 70%. This gave a final pancake 39 mm high and 170 mm in diameter.

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The microstructure after forging was approximately 50 μm equiaxed-gamma (TiAl) grains with about 5 μm alpha-2 (Ti_3Al) particles at the grain boundaries (figures 1(a) and (b)).

Specimens 4x8x60 mm were sectioned from the forged pancake by electrical discharge machining. Before heat treatment, specimens were wrapped in tantalum foil and encapsulated in evacuated silica ampoules at a vacuum of 10^{-6} torr and back-filled with argon. Heat treating at a temperature of 1390°C for 4 hours produced a coarse fully lamellar colony size of approximately 3 mm (figure 1(c)), while heat treating at a temperature of 1420°C for 2 minutes produced a finer fully lamellar colony size of about 0.8 mm (figure 1(d)). Prior to testing, the specimen surfaces were polished to a 1 μm surface finish.

A notch was cut in the centre of the specimen to a depth of 2 mm using a fine slow speed diamond saw. A sharp precrack (of depth typically 200-500 μm) was introduced by fatigue loading the specimen in uniaxial compression at an R ratio of 0.1 (where R is the ratio of the minimum stress to the maximum stress applied over the fatigue cycle). Since this type of fatigue precrack is indistinguishable fractographically from the subsequent fracture surface it was necessary to heat tint the precracked specimen at a temperature of 600°C for 15 minutes. Fracture toughness tests were performed in four point bend at room temperature on an ESH servohydraulic test machine. A clip gauge was attached to the specimens using knife edges to determine load-displacement behaviour. Nine tests were carried out on the two lamellar microstructures and four tests on the as-forged microstructure.

RESULTS AND DISCUSSION

K_Q values have been determined from the load-displacement graphs by use of the 5% offset procedure [2], while K_{max} values were calculated from the maximum load. Since plane strain conditions are not met then these K_Q values can not be strictly regarded as K_{IC} values. The results are given in the following table:

Microstructure	No. of Tests		K_Q or K_{max} (MPa√m) Average	High	Low	Std. Dev.
As-Forged	4	K_Q	3.9	4.8	3.0	1.0
		K_{max}	5.0	7.3	3.2	1.9
Fine Fully Lamellar	9	K_Q	9.0	10.5	6.1	1.3
		K_{max}	15.2	19.3	12.0	2.6
Coarse Fully Lamellar	9	K_Q	7.7	9.7	4.6	1.7
		K_{max}	16.8	22.3	13.6	2.8

Clearly the as-forged microstructure has poor fracture toughness both in terms of K_Q and K_{max} values. The low toughness of 4 MPa√m is due to the poor resistance to crack growth of the equiaxed-gamma microstructure. At room temperature, crack propagation occurs by mainly transgranular cleavage (see figure 3(a)) with some intergranular decohesion. Both are brittle failure mechanisms and offer little resistance to crack propagation.

Heat treating to produce a fully lamellar microstructure gives a relatively large improvement in K_Q, but the toughness is still relatively low compared to other materials such as conventional Ti alloys (between 30-100 MPa√m [3]). Values of 9.0 MPa√m and 7.7 MPa√m for the fine and coarse fully lamellar microstructures respectively are approximately twice that of the as-forged material. However, there appears to be slightly greater scatter in these results compared to the equiaxed-gamma material. The average K_{max} values of 15.2 MPa√m for the fine fully lamellar and 16.8 MPa√m for the coarse fully lamellar microstructure are approximately twice the calculated average K_Q values. The K_{max} values are somewhat similar to those obtained previously for as-cast Ti-48Al-2Mn-2Nb heat treated to give a fully lamellar microstructure [4].

The stress intensity-displacement curves for the fully lamellar microstructures are shown in figure 2 (here, load has been converted to stress intensity without including any stable crack growth prior to final fracture). A high K_{max} value curve has no indication of a pop-in event occurring prior to maximum load and since these are caused by interlamellar splitting [4] then the crack must be propagating by a predominantly translamellar mechanism. The fracture surfaces of the high K_{max} specimens do indeed show there to be a large proportion of translamellar crack propagation directly under the precrack, see figures 4(i). A low K_{max} value curve tends to show pop-in behaviour prior to maximum load which indicates that interlamellar splitting has occurred. Again the fractographs (figures 4(ii)) show this to be the case.

Fracture surfaces of the precrack boundary area of the fully lamellar specimens giving average, high and low K_{max} values are shown in figures 3(b) and 4. The coarse fully lamellar microstructure clearly shows the importance of the lamellae orientation to the mode I crack direction on fracture toughness. A high toughness (figure 4(b)(i)) is due to the large amount of translamellar crack growth, note that the precrack boundary is approximately 100-200 μm from the notch root. A low toughness (figure 4(b)(ii)) is caused by the large amount of interlamellar splitting below the precrack boundary. An average toughness (figure 3(b)(ii)) has a mixture of trans- and interlamellar failure. For the fine fully lamellar microstructure the same trend is observed with a high toughness value having a large proportion of translamellar failure (figure 4(a)(i)), a low value showing interlamellar splitting (figure 4(a)(ii)), and an average value showing both inter- and translamellar failure. However, the same trend does not appear to follow for the K_Q values. For example, figure 4(b)(i) shows a large proportion of translamellar crack propagation, but has a low K_Q of only 6.1 MPa√m.

Comparing the two fully lamellar microstructures it is seen that the average and the range of K_Q and K_{max} values are similar, hence there appears to be little effect of colony size on fracture toughness for the range 800 to 3000 μm. This is surprising since the coarse lamellar material has a colony size similar to the breadth of the specimen, hence only one or two grains will be sampled beneath the precrack. However, the fine lamellar microstructure should have at least five colonies being sampled under the precrack. It may simply be that these specimens are too thin and the probability of finding five colonies similarly aligned in the fine fully lamellar microstructure is relatively high. This would explain the similar distribution in toughness values. Alternatively fracture may be dominated by, for example, three of the five grains under the precrack. If these fail by interlamellar splitting then the toughness will be low and if they fail by translamellar fracture then it will be high. Future work on thicker test specimens with the same colony sizes used in this work should clarify the situation.

There is a greater actual spread in the K_{max} values compared to the K_Q values (8 and 4 MPa√m respectively) for the fully lamellar microstructures. But as a percentage of average values there is little difference because the average K_{max} value is about twice the average K_Q value. The similarity in percentage spread of K_Q and K_{max} values suggests that the variation is controlled in both cases by the amounts of trans- and interlamellar fracture just ahead of the crack. However, there are large differences in the ratio of K_{max}/K_Q for each specimen. This ratio varies between 1.53 and 3.67 for the coarse fully lamellar, 1.23 and 2.16 for the fine fully lamellar, and 1.06 and 1.54 for the as-forged specimens. The ratio is generally higher for the fully lamellar microstructures and is probably due to the "toughening" effect of translamellar crack propagation, increased crack deviation from the mode I direction due to interlamellar splitting and the large colony size. The lower values, and tighter spread, for the as-forged material is due to the fine 50 µm grain size (hence small crack deviations) and the predominant brittle transgranular cleavage fracture mode.

CONCLUSIONS

The gamma-based titanium aluminide material studied here has a low fracture toughness. The as-forged equiaxed-gamma material has the lowest toughness K_Q values of 4 MPa√m, whilst higher values of between 8-9 MPa√m are achieved for the two fully lamellar microstructures. Varying the colony size from 800 to 3000 µm appears to make little difference to fracture toughness, although this may be due to the small number of grains sampled by the fatigue precrack in these relatively thin specimens. The K_{max} values are about twice the calculated K_Q values for the fully lamellar microstructures.

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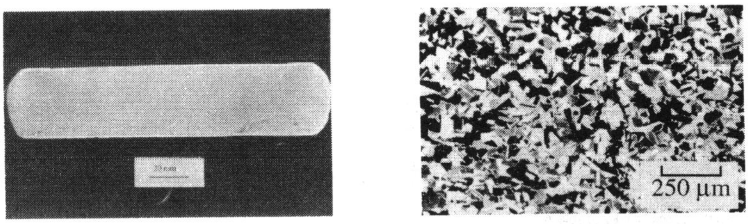
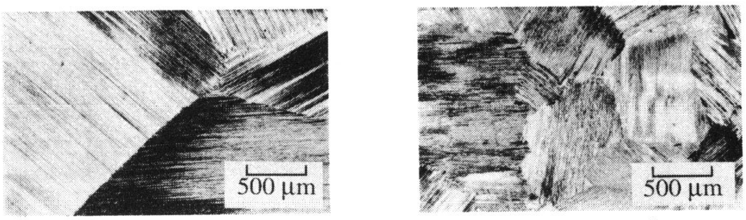
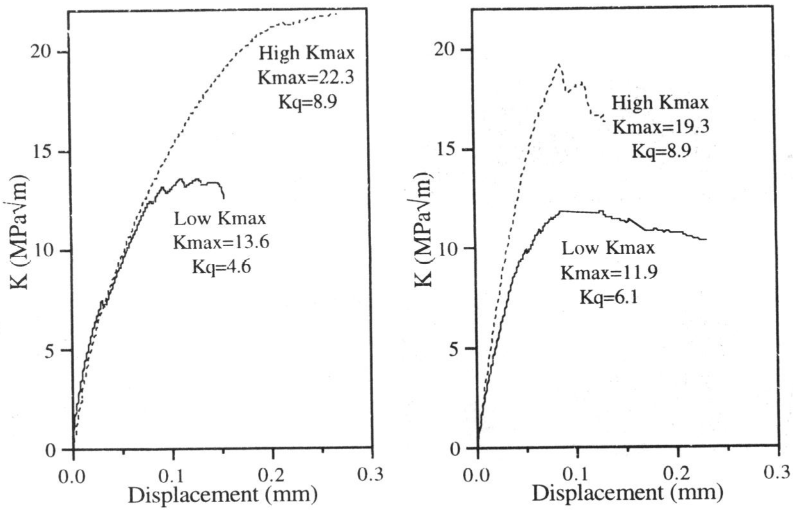


(a) As-forged macrostructure. (b) As-forged microstructure.



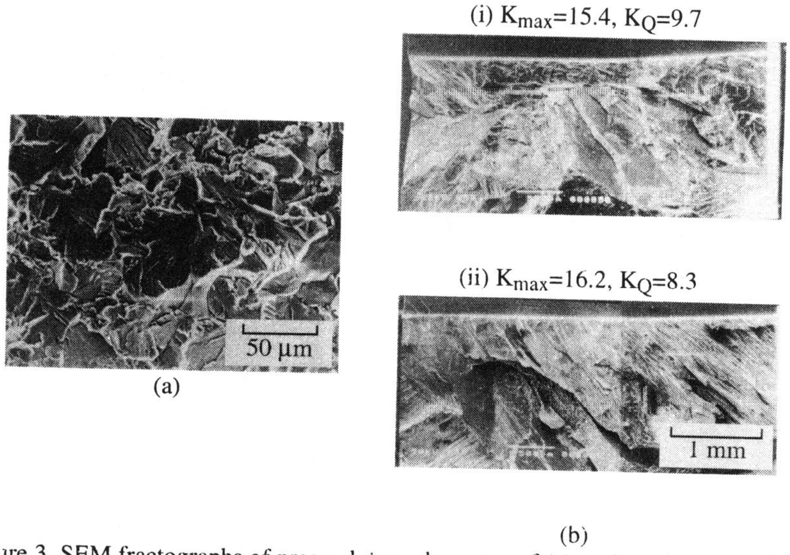
(c) Coarse fully lamellar. (d) Fine fully lamellar.

Figure 1 As-forged (a) macrostructure and (b) microstructure and the two fully lamellar microstructures (c, d).



(a) Coarse Colony Fully Lamellar (b) Fine Colony Fully Lamellar

Figure 2 Stress intensity versus clip gauge displacement for high and low K_{max} of coarse and fine fully lamellar specimens.



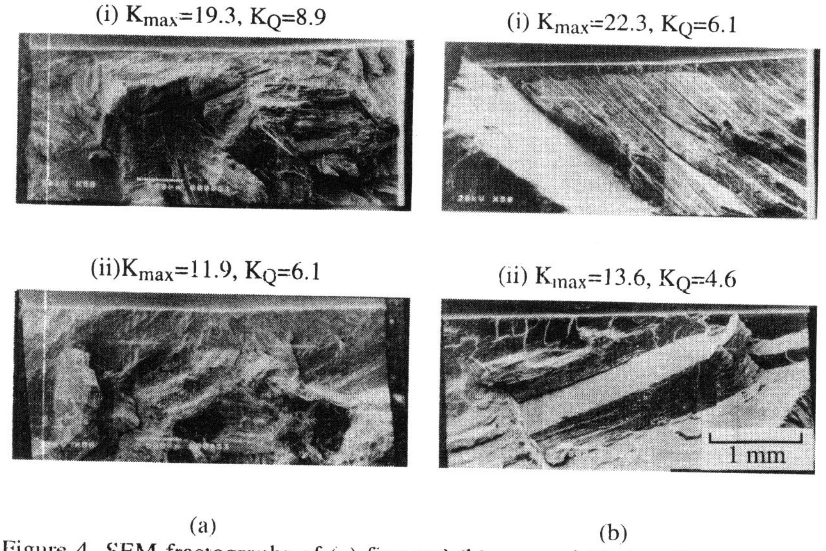
(i) K_{max}=15.4, K_Q=9.7

(ii) K_{max}=16.2, K_Q=8.3

(a)

(b)

Figure 3 SEM fractographs of precrack boundary area of (a) as-forged and (b) (i) fine and (ii) coarse fully lamellar giving average K_{max} values.



(i) K_{max}=19.3, K_Q=8.9

(i) K_{max}=22.3, K_Q=6.1

(ii)K_{max}=11.9, K_Q=6.1

(ii) K_{max}=13.6, K_Q=4.6

(a)

(b)

Figure 4 SEM fractographs of (a) fine and (b) coarse fully lamellar specimens giving (i) high and (ii) low K_{max} values.