Micromechanisms of fracture have been studied in experimental alloys based on Ti₃Al (α₂) based titanium aluminides, with and without the addition of molybdenum (Mo). At ambient temperature, failure occurs by transgranular cleavage. The critical failure event appears to be the fracture of individual α₂ laths in the base alloy, but of a group or packet of laths in the Mo containing alloy. Thus, although the lath size is refined by the addition of Mo, the fracture toughness is not increased. At the test temperature of 500°C, the base alloy fails by plastic collapse, and failure occurs by a process of micro-void coalescence. In contrast, the Mo containing alloy fails in a macroscopically brittle manner, and by a process of intergranular micro-void coalescence.

Introduction

In recent years titanium aluminides have shown good potential for use within the aerospace industry as low density materials possessing high temperature strength [1]. The drive behind the development of these materials comes from the considerable weight advantages to be gained if titanium alloy or nickel-based superalloy components within a gas turbine engine can be replaced by low density materials with comparable mechanical properties. The reduction in engine weight combined with the increase in operating temperatures that titanium aluminides offer would assist in the development of engines with greatly increased speed and efficiency.

The ordered intermetallic compounds Ti₃Al (α₂) and TiAl (γ) have received considerable attention over the past two decades due to their good strength and modulus retention at high temperatures. However, both these compounds suffer from poor low temperature ductility and poor fracture toughness. The addition of niobium (Nb) to Ti₃Al has a significant effect on the ductility [2] due both to the stabilisation of the high temperature β phase and to a reduction in the planarity of slip. Commercial alloys based on Ti₃Al with Nb additions, such as Ti-24Al-11Nb and Ti-25Al-10Nb-3V-1Mo, have been identified as possessing a good balance of elevated temperature rupture strength and room temperature ductility.

Oxidation resistance currently limits the potential use of α₂ based alloys to the compressor section of gas turbines where titanium alloys are currently used. Alloys based on the γ phase show greater promise at higher temperatures as replacement materials for superalloys, but suffer severely at present from a lack of room temperature ductility or...
toughness. Alloys based on α2 are at a more advanced stage of development than γ alloys, and fabricated components have successfully undergone engine trials [3]. The use of α2 alloys is also being considered as a matrix material for advanced composites. This paper forms part of a fundamental study into the mechanisms of fracture of a range of experimental alloys based on Ti3Al.

**Experimental**

The materials used in the work presented here form part of an alloy development programme being carried out at RAE Farnborough [4]. Extensive work on alloys with a range of aluminium and niobium contents identified a base alloy composition of Ti–23Al–11Nb–0.9Si (atomic %) as showing potential for reasonable room temperature elongation (up to 3%) and excellent creep rupture strength (in excess of 1000 hours at 625°C and 250MPa stress). The further substitution of molybdenum (Mo) in this base alloy, resulting in the composition Ti–23Al–9Nb–2Mo–0.9Si, has shown increased high temperature strength and creep resistance [5]. For the purpose of this present study these two alloys underwent thermo-mechanical processing in the form of a ~-extrusion followed by controlled heat treatments in the β phase field. Three different alloy conditions have been studied: the base alloy was solution treated at 1150°C for 1 hour followed by two different ageing treatments (625°C for 2 hours, or 800°C for 1 hour), and the Mo alloy was solution treated at 1150°C for 1 hour followed by an age at 800°C for 1 hour.

Single edge notched (SEN) specimens of dimensions 50x10x10mm were cut from lengths of 50mm diameter extruded bar by electrical discharge machining. Each specimen was notched to a depth of 2.5mm using a 150µm thick diamond blade. In addition a small number of notched bend specimens of a specific notch profile [6] were also machined from the base alloy in the 625°C aged condition. Notches were ground to the required notch root radius of 200µm and notch included angle of 45°. SEN bend specimens were pre-cracked to an a/W of 0.45–0.55 using an Amsler Vibrophore electro-magnetic resonance machine as described elsewhere [7]. Fracture toughness testing was carried out at room temperature in accordance with BS 5447:1977[8], using a four point bend geometry on an ESH servo-hydraulic testing machine. At elevated temperatures the constraints of the fixtures necessitated the use of three point bending, and toughness values were derived solely from a cross-head displacement versus load trace and so do not qualify as $K_{IC}$ values.

Fractographic examination was carried out on an ISI 100A scanning electron microscope (SEM), operating at 0° tilt, 20kV. Additionally, some metallographic sections cut transverse to fracture surfaces were also examined.

**Results**

Optical and SEM micrographs of Ti–23Al–11Nb–0.9Si and Ti–23Al–9Nb–2Mo–0.9Si in the β heat treated conditions are shown in Figure 1. These show the 'basket weave' microstructures which result from air cooling from the single phase β field. Qualitative energy dispersive X-ray (EDX) analysis on the 625°C aged condition has shown the thin regions surrounding each lath (the light regions of Figure 1b) to be richer in niobium but to have a lower aluminium content compared to the centre of the laths. Examination of foils in a transmission electron microscope (TEM) has shown this surrounding phase to be retained β. The microstructures of the two ageing conditions of the base alloy appear to be basically the same, see Figure 1c), but Mo has the effect of refining the 'basket-weave' structure significantly, Figure 1d).

The average values of fracture toughness exhibited by the alloys under investigation are given in Table 1. The room temperature values of the base alloy are relatively low, with marginally the best toughness exhibited by the alloy aged at 625°C. The addition of Mo decreases the toughness of the alloy. The SEM micrographs in Figure 2 show the fracture surfaces of specimens tested at room temperature. Fracture occurred by transgranular cleavage in all conditions. Moreover, at high magnification, for the base alloy, see Figure 2a), river lines can be seen clearly within an individual α2 lath. Thus for the base condition, the failure of
an individual lath appears to be a discrete failure event. This is in contrast to the fracture of the Mo containing alloy where cleavage appears to radiate out through a group of individual laths, Figure 2b). Ductile tear ridges are present around the cleavage facets on the surfaces of both base alloy conditions, but appear less prevalent on the fracture surface of the Mo containing alloy. The appearance of the fracture surface of the Mo containing alloy is also smoother with a larger apparent cleavage facet size.

Table I - Fracture toughness results.

<table>
<thead>
<tr>
<th>Alloy Condition</th>
<th>Room temperature $K_{IC}$ (MPa(\sqrt{m}))</th>
<th>500°C - Nominal toughness (MPa(\sqrt{m}))</th>
</tr>
</thead>
<tbody>
<tr>
<td>Base Alloy, $\beta+625°C$ age</td>
<td>15.7</td>
<td>39.3</td>
</tr>
<tr>
<td>Base Alloy, $\beta+800°C$ age</td>
<td>14.5</td>
<td>35.9</td>
</tr>
<tr>
<td>Mo Alloy, $\beta+800°C$ age</td>
<td>12.6</td>
<td>44.9</td>
</tr>
</tbody>
</table>

At 500°C the base alloy conditions fail by a process of plastic collapse. The fracture surfaces in Figure 3a),b) show extensive micro-void coalescence and large featureless areas of localised shear. Fracture in the Mo alloy at this temperature is intergranular, as shown in Figure 3c). The intergranular failure mechanism is one of ductile micro-void coalescence rather than cleavage as is evident from Figure 3d). The load/displacement curves for the base alloy and the Mo alloy in the same condition are shown schematically in Figure 4. Failure in the Mo alloy was characterised by a series of 'pop-ins' (determined from a sudden reduction in load), whereas in the base alloy the failure occurred by plastic collapse. The values of toughness at 500°C quoted in Table I are nominal since the use of a peak load/cross-head displacement trace rather than a clip gauge at elevated temperatures does not provide a valid $K_{IC}$ under the conditions given in BS 5447. Also, due to the substantial plasticity exhibited at
elevated temperatures, it would be expected that the use of elastic-plastic fracture mechanics parameters to characterise failure would result in an increased level of toughness for the base alloy conditions.

Notched bend tests on the base alloy in the 625°C aged condition gave an average cleavage fracture stress, \( \sigma_F^* \), value of 1280MPa at room temperature which is a little over twice the yield stress. Methods of evaluating \( \sigma_F^* \) have been described in detail elsewhere, see for example [9].
Failure of these alloys at room temperature occurs by a process of transgranular cleavage. From the observation of secondary microcracks behind the main crack front in the sectioned fracture surfaces (Figure 5), fracture at room temperature in the base alloy would appear to proceed by cleavage across α₂ laths followed by tearing of the more ductile intervening β phase. Ligaments of this β phase can be seen left bridging some microcracks, providing "crack tip shielding". Similar observations have also been made in other studies on α₂ based alloys [10-12].

It is accepted generally that in many body centred cubic metals transgranular cleavage occurs under a critical value of tensile stress. Moreover, some recent work on "super α₂" in an α₂+β heat-treated condition has shown that the cleavage fracture stress, $\sigma_F^*$, is independent of temperature [13], indicative of a critical tensile stress controlled fracture criterion. Such failures are associated with a growth controlled mechanism where a microcrack propagates catastrophically at a critical value of tensile stress ($\sigma_F^*$). If a growth controlled process is assumed in this case, then critical microcrack size can be linked to the cleavage fracture stress through a modified Griffith relationship:
\[ \sigma_r = \sqrt{\frac{4E\gamma_p}{\pi(1-\nu)d}} \]

where \( E \) is Young's modulus, \( \gamma_p \) is the effective surface energy, \( \nu \) is Poisson's ratio, and \( d \) is the critical microcrack size. Values can be assigned to most of the variables in this equation for the base alloy 625°C aged condition (which has been the subject of the most extensive study so far \[14\]). Measurements of cleavage fracture stress are in the range 720–1280MPa, with values being calculated from uniaxial tensile results and notched bar tests in pure bending respectively. The modulus is taken as 130GPa and Poisson's ratio as 0·26\[15\]. Estimation of the effective surface energy is, however, more problematic. Values for bcc metals range from 2–20Jm\(^{-2}\) \[16\]. It is likely that a brittle hexagonal intermetallic will exhibit a value towards the bottom of this range, and so an estimate of between 2–20Jm\(^{-2}\) has been made for \( \gamma_p \), and would not seem too unreasonable. With these values applied, microcrack sizes which would result in the fracture stresses observed are of the order of 0·2–7µm. Although this range is relatively large, it should be noted that these values are estimated outer limits. Even so, the microstructural feature of the \( \beta \)-heat treated base alloy which corresponds most closely in order of magnitude to this critical microcrack size is the \( \alpha_2 \) lath width.

This analysis, although only of a preliminary nature, when combined with the observations of cleavage across \( \alpha_2 \) laths made on sectioned fracture surfaces, Figure 5, and the river markings on individual \( \alpha_2 \) laths, Figure 2a), would suggest that the measured fracture stress is consistent with the propagation of a crack across a single \( \alpha_2 \) lath being the controlling factor in the fracture of this alloy at room temperature. If this is so, then since a smaller microcrack size leads to an increased fracture stress, it should thus be possible to raise this fracture stress and improve the toughness of the alloy by reducing the lath width size. It has indeed been reported elsewhere \[17\] that a decrease in the size of the \( \alpha_2 \) phase can result in substantial increases in room temperature tensile strength, fracture strength, ductility, and fracture toughness. The addition of Mo to the base alloy has the effect of significantly reducing the lath width size and keeping yield stress values the same, yet as can be seen from Table 1, this does not appear to improve the fracture toughness of this alloy. Most important, fractographic observations suggest that after the addition of Mo, the critical failure event is now the fracture of a group or packet of \( \alpha_2 \) laths, see Figure 2b). This new unit size appears to be somewhat larger and hence remains consistent with the use of a modified Griffith equation. The precise role of the Mo is yet to be determined but it may plausibly change the nature of the surrounding \( \beta \) phase. Indeed bridging ligaments such as are observed in the base alloy have not yet been observed in the Mo alloy. Analytical TEM will be used to confirm this suggestion.

![Figure 6 - SEM micrographs of the sectioned fracture surface at 500°C of a) base alloy aged at 625°C, b) Mo containing alloy.](image)

At a test temperature of 500°C the base alloy is above its brittle to ductile transition and failure occurs primarily by a process of micro-void formation, growth and coalescence. Void formation would appear to occur at the boundary between the \( \alpha_2 \) and \( \beta \) phases as revealed in the sections through fracture surfaces shown in Figure 6a). The Mo alloy appears to be
macroscopically brittle at 500°C, see Figure 4. In section, the fracture surface suggests that failure occurs along prior β grain boundaries, see Figure 6b), and this is confirmed fractographically, Figure 3c). However, the failure micromechanism is one of intergranular micro-void coalescence, see Figure 3d). These voids are of the order of 1µm and are relatively shallow. The precise reason why the addition of Mo promotes intergranular failure at 500°C is at present unclear. Further work is required to understand the fracture behaviour of this alloy over this range of test temperatures.

Conclusions

1. Fracture at room temperature occurs by transgranular cleavage. The room temperature fracture toughness is relatively low in all alloy conditions studied to date. Ageing temperature appears to have little effect on the value of fracture toughness or the mechanism of failure in the base alloy. It is postulated that the mechanism of failure in the base alloy can be characterised by microcrack initiation within individual α2 laths and their subsequent propagation through more ductile β regions to give catastrophic failure. The addition of Mo refines the lath structure but does not promote increased toughness. Fractographic observations suggest that the critical failure event is now the simultaneous failure of a group or packet of α2 laths.

2. At the test temperature of 500°C the base alloy is above its ductile-brittle transition and failure occurs by a process of micro-void initiation, growth, and coalescence. Void initiation appears to occur at the boundary between the α2 and β phases. The Mo containing alloy fails in an intergranular manner at this temperature with void formation and growth occurring along prior β grain boundaries.

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References


