A REVIEW OF FACTORS INFLUENCING THE DUCTILITY AND FAILURE IN SOME Ti AND TiAl ALLOYS

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ABSTRACT

The factors which influence the ductility and fracture behaviour of some beta Ti alloys and high strength TiAl alloys have been investigated. In the case of the beta Ti alloys it has been shown that small alloying additions can either improve the ductility or result in embrittlement. The different mechanisms responsible for these changes are discussed. The failure in tension and during fatigue of high strength TiAl-based alloys has been shown to be associated with plastic flow, which occurs well below the 0.2% proof stress in these alloys. This is discussed in terms of the large variation of yield stress for differently oriented crystals in polycrystalline fully lamellar samples.

KEYWORDS

Ti Alloys, TiAl, microstructure and crack initiation, acoustic emission.

INTRODUCTION

The results obtained in two separate pieces of research on Ti-based alloys will be discussed in this paper; they are linked by the fact that the work illustrates the complex ways in which microstructure can influence the ductility and fracture behaviour of these alloys.

A new beta Ti alloy has recently been developed[1] in the IRC in response to the requirement that a low cost burn-resistant alloy was required by aero engine manufacturers. Pratt and Whitney [2] had developed Alloy C, a high vanadium, high chromium alloy, which was burn-resistant, but was very expensive. The use of master alloys, to introduce V and Cr, which contained aluminium allowed the cost of the alloy to be reduced by a factor of ten, but the addition of these master alloys resulted in a totally brittle alloy [3]. The work described in this paper traces the development of the alloy Ti25V15Cr2Al0.2C (wt%) which has excellent ductility. In addition despite the addition of 2%Al it is stable against precipitation of alpha (which leads to embrittlement) during high temperature exposure.

Alloys based on TiAl have been the subject of an enormous amount of research over the last fifteen or so years and alloys, which have yield strengths of about 1000MPa, have been developed, which show acceptable ductility, although this is typically only about 1- 2% at room temperature[e.g. 4,5, 6]. The preferred structure for these alloys is one in which the two phases alpha 2 (Ti_3Al) and gamma (TiAl) form a lamellar structure. The factors which increase the strength of these alloys are mainly refinement of the grain size and of the interlamellar spacing and the level of the Al content; the lower Al alloys are stronger since they contain more Ti_3Al which is far stronger than TiAl. Earlier work has shown that dislocations in the alpha 2 contributed to the general plasticity of polycrystalline samples, but it was clear that the local stress required to activate this slip, especially when it involved dislocations of b = 1/6<11~26>, was very high[7,8]. Failure of these alloys has been shown to be associated with cracking between the lamellae of alpha 2 and gamma, which occurs if slip transfer between the alpha 2 and gamma requires too high a local stress.

In some recent work[9,10] we have found significant acoustic emission well below yield, as well as during and after yield in polycrystalline TiAl alloys. The work summarised here has been carried out in order to understand the origin of these signals and to assess their significance. The availability of samples of fully lamellar Ti44Al8Nb1B, which have extraordinarily reproducible properties, has allowed samples to be tested at stress levels which are known to be at defined levels below the 0.2% yield stress. The experiments reported in this paper cover work aimed at interpreting the acoustic emission signals generated at such low stresses and to assess the role any cracks, which are induced by loading to a defined stress level, play in subsequent failure in fatigue.

EXPERIMENTAL

The ingots for both alloys were double plasma melted after which they were extruded and heat treated to produce fine grained samples. In the case of the Ti25V15Cr2Al0.2C the presence of the carbide precipitates limited grain growth and a grain size of about 90µm was obtained. In the case of the Ti44Al8Nb1B alloy the grain size after heat treatment was about 70µm and the microstructure was fully lamellar, consisting of parallel plates of TiAl and Ti_3Al.

Samples of each alloy were cut and polished for microstructural examination. Some of the Ti25V15Cr2Al0.2C alloys samples were stress-relieved at 600 for 2h, followed by exposure at temperatures between 450 to 550°C for 24h to 1000h. All the samples were cut, polished and etched for conventional optical microscopy and scanning electron microscopy (SEM). Transmission electron microscopy (TEM) specimens were prepared by either ion beam thinning or twin jet polishing. TEM was carried out on either a JEOL 4000FX or a Phillips CM20 transmission electron microscope interfaced with LINK EDX analytical systems.

Tensile tests on Ti25V15Cr2Al0.2C were carried out using a Zwick screw-driven test machine. Tested specimens were all sectioned longitudinally and polished for optical and SEM examination to study the microstructure close to the fracture surface. An Instron 1273 machine with a 100 kN load cell was used for all the tensile and fatigue tests on Ti44Al8Nb1B. Tensile tests and fatigue tests were carried ratio of R=0.1 ($R=\sigma_{min}/\sigma_{max}$, here σ_{min} and σ_{max} are the minimum and maximum applied stress, respectively), and a frequency of 10 Hz.

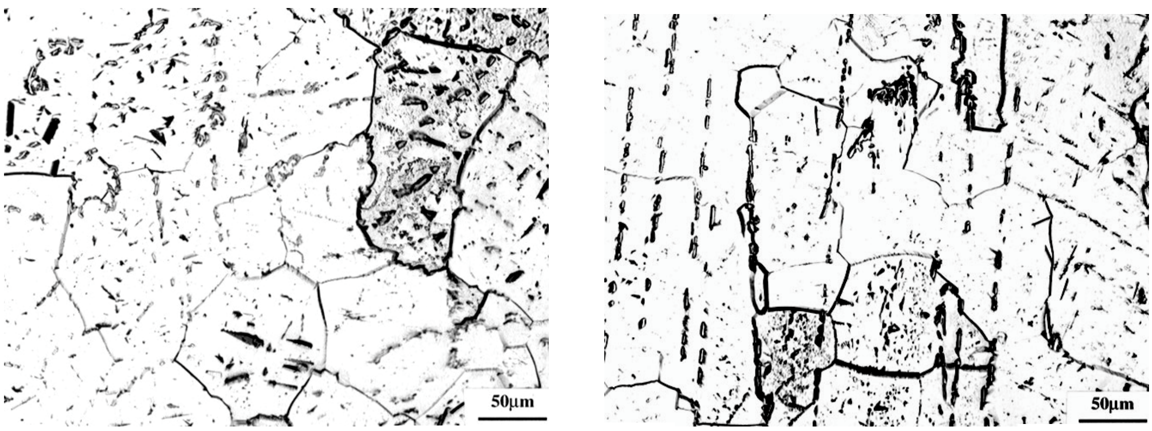
In both tensile and fatigue tests, two AE sensors were glued to the ends of the gauge length. Acoustic emission events detected during testing were amplified by a pre-amplifier and analysed by a Mistras (Physical Acoustic Co.). The gain of the pre-amplifier was 40 dB. The threshold was 45 dB, which meant that events of amplitude smaller than 45 dB were not recorded.

RESULTS AND DISCUSSION

Beta Ti alloys based on Ti25V15Cr

As noted above the addition of Al (as a constituent of the VCr master alloys) resulted in the ductile base alloy of Ti25V15Cr becoming totally brittle. It was found that two factors underlay this embrittlement. The first was the fact that the master alloys could contain high oxygen levels, which increased the level in the alloy to over 1500ppm, but more fundamentally the alloy became ordered. This was clearly evident in alloys containing about 6%Al where B2 superlattice maxima were visible in electron diffraction patterns and anti phase domains could be imaged. The dislocations introduced by deformation were arranged as pairs and they also generated intense slip bands giving rise to the low ductility. At lower Al contents diffraction cannot reveal order (because of the similar scattering factors of the atoms occupying the two sites in this structure). Surprisingly even neutron diffraction failed to reveal superlattice maxima but when the relative intensities of fundamental and superlattice reflections were calculated (from a knowledge of site occupancy) it became clear that indeed the predicted intensity of superlattice maxima would drop to zero even if the alloy were perfectly ordered. Electron diffraction was the most sensitive of all diffraction techniques used, but the dislocation arrays in deformed samples and data obtained using calorimetry on well annealed samples, showed that even at 2%Al the alloy was weakly ordered [3].

At an Al content of 2% it was shown that good ductility could be achieved at oxygen contents of about 800ppm, but at this oxygen level the alloy is expensive and re-cycling would not be practicable. Hence further alloy development work was carried out in which it was shown that addition of 0.2C resulted in excellent ductility, even at oxygen contents as high as 1500ppm. The improvement in ductility through addition of C was shown to be associated with the fact that the TiC which is formed is non-stoichiometric, Ti_2C, and that this carbide acts as a getter (removing most of the oxygen from the matrix) to form Ti(CO). The TiC particles also act as a very effective grain refiner when the alloy is thermomechanically processed since they limit grain growth. The microstructure of as-extruded samples consists of equiaxed _ grains and titanium carbides in both cross and longitudinal sections. Carbides are aligned along the extrusion axis. The carbides are rods with an aspect ratio ranging from 2 to 5 and a length of 5-15 µm, as illustrated in Fig 1.



(a) 50µm (b) 50µm

Fig. 1. Optical micrographs showing as-extruded microstructures of burn resistant alloy Ti-25V-15Cr-2Al-0.2C-0.13O. (a) transverse section, (b) longitudinal section.

The role of the carbides was found to be even more far reaching. Alloys containing Al additions became even more brittle on exposure to likely service temperatures of 450 to 550°C. This brittleness is illustrated in fig 2 and was shown to be due to the precipitation of alpha at the beta grain boundaries. When carbides are present the ductility does decrease from about 20% to about 10% but the amount of grain boundary alpha is dramatically reduced. This reduction occurs firstly, because oxygen is an alpha stabiliser and the tendency to form alpha is reduced because the oxygen has been taken up by the carbides. Secondly, because the much smaller grain size in the carbon-containing alloy increases the total grain boundary area the mean grain boundary concentration of alpha would be decreased even if the same amount of alpha were precipitated. Finally, the carbides themselves act as heterogeneous nucleation sites for the alpha, again reducing the grain boundary precipitation.

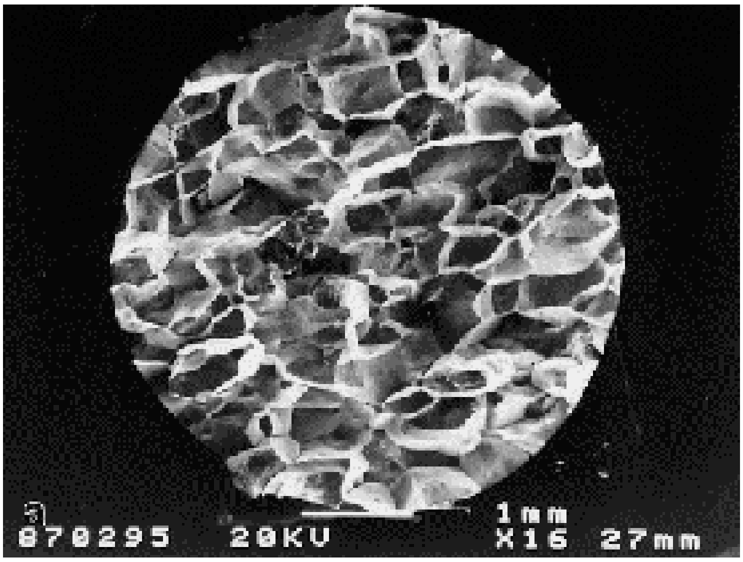


Fig 2 Secondary electron scanning electron micrograph of the fracture surface of a sample of Ti25V15Cr 4Al after exposure at 550°C for 500h.

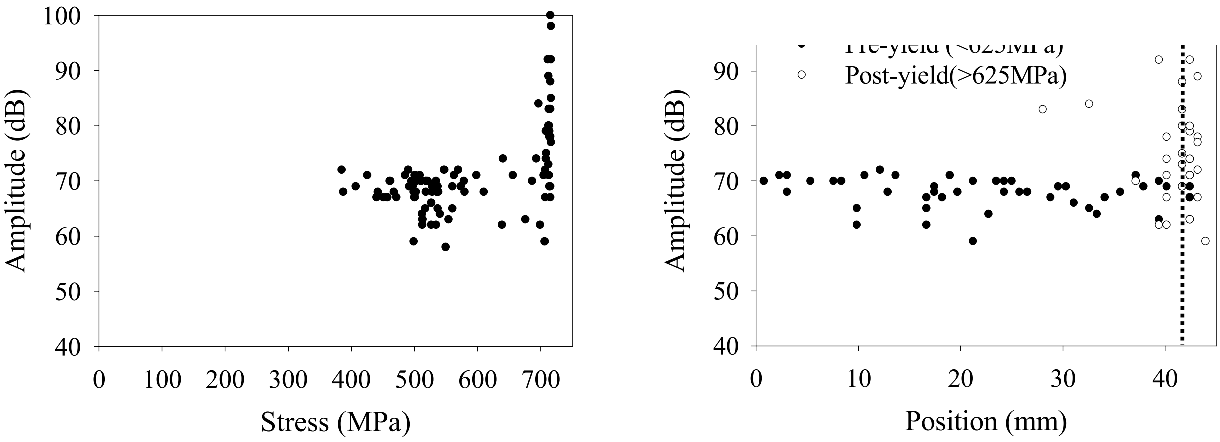
It is thus clear that the important factors that limit ductility in Ti25V15Cr(2-6)Al are the oxygen content, the grain size (which is very large in the absence of carbides, because grain growth is rapid) and the Al content. In exposed samples the tendency to precipitate alpha at grain boundaries, is increased with increase of oxygen and of Al and is more potent in causing embrittlement in coarse grained samples. The selection of a maximum Al content of 2wt% together with the addition of carbon to form a fine distribution of carbides effectively overcomes all of these factors. It is highly likely that improvements in properties could be achieved with other similar alloys and work is underway to investigate this possibility.

Tensile tests on Ti44Al8Nb1B

Two series of experiments have been carried out on Ti44Al8Nb1B in order to understand the significance of the acoustic emission events referred to in the introduction. In the first experiments samples were loaded in tension and the acoustic events correlated with any changes in the defect structure, including the presence of cracks. In the second series of experiments, samples were fatigued in tension-tension either after pre stressing to 600 MPa (note the 0.2% proof stress is 625MPa for this alloy) or fatigued without pre stressing at maximum stress levels of between 380MPa and 500MPa.

Figure 3 shows a typical AE result obtained during tensile loading of the Ti44Al8Nb1B alloy. It is obvious that a large number of events can be detected. It can be seen, Fig3 that most of the AE events occur above 400 MPa. Those events have a range of amplitudes of 60-100 dB and are uniformly distributed within the gauge length. Surprisingly some events have been detected at stresses as low as 100 MPa, and the number of AE events detected at low stresses varies with individual samples. A large number of samples have been examined using acoustic emission and in all cases significant emissions were detected at stresses of about 400MPa and above.

Some of the tensile tests were interrupted at low stresses, as soon as one or two AE events were detected and cracks were found under optical and SEM examination. Cracking appears to take place between lamellae within a colony and more than one crack was often observed. Importantly the observations showed that acoustic events were always associated with visible cracking and that cracking occurred at much lower stresses than expected. There was no obvious tendency for the surface cracks to lie along any specific directions; some were nearly perpendicular to the stress axis, but some were clearly inclined to the axis. The cracking associated with acoustic signals, which occurred at stresses as low as 100MPa was associated with cracks on the sharp corners of the test-pieces and there is little doubt that the use of the rectangular-shaped samples has led to events at such low stresses.



(a) (b)

Fig. 3 (a) AE events as a function of stress level generated in the Ti44Al8Nb1B during stressing to 700 MPa and showing that most of AE events occur at stresses above 400 MPa; (b) showing the distribution of AE events along the gauge length during pre- and post-yielding. The dashed line indicates the position where the sample failed.

Extensive examination by TEM has been carried out, which showed that in this alloy twinning was not associated with acoustic emission at the stress levels investigated, i.e. up to the macroscopic yield stress. Because the Ti44Al8Nb1B alloy contained hard boride particles, which may have cracked during tensile straining, a large number of these were examined in TEM in deformed samples. No obvious fracture was observed of the borides at the low strains used.

The significant and very surprising result, which has been obtained is that cracking occurs at stresses well below the 0.2% proof stress of Ti44Al8Nb1B.The samples were carefully examined before stressing and after polishing and/or etching and in no case were any surface cracks observed. Over 200μm was removed from the surfaces during polishing, so that any residual stresses (as well as any cracks) associated with the initial spark machining will have been removed.

The plastic deformation of lamellar TiAl samples has been studied by Inui et al [11] and in this work they have shown that the yield stress is greatest when the stress axis is either parallel to or perpendicular to the lamellae interface. Plastic deformation is able to occur in polycrystalline samples at applied stresses significantly below the macroscopic yield point because TiAl samples are made up of these soft and hard colonies (or grains). The fact that dislocations are able to move through the soft grains means that when these dislocations encounter a grain where the lamellae are oriented in such a manner as to make slip transfer difficult, they will be stopped and will generate a high stress in that grain. As the imposed stress increases, further plastic strain will occur and eventually the stress in these grains can reach levels above the fracture stress, so that fracture will occur.

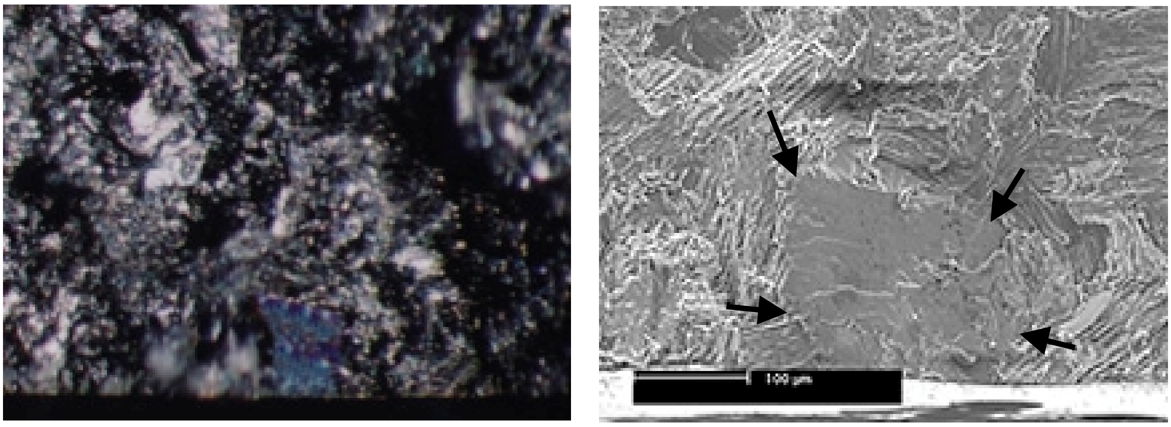
Fatigue testing of pre-stressed and un pre-stressed samples. The influence of these precracks on fatigue life has been studied using tension-tension fatigue tests at maximum stresses of 380, 440, 500MPa . Some samples were tinted after the tensile test so that any cracks present at that stage could be identified on the final fracture surface. Some fatigue tests were interrupted and the specimen tinted and the test continued.

The maximum cycles to failure of the test-pieces are listed in the Table. It is clear from the observations that at a stress of 380 MPa, tests always run out even when the sample has been pre-stressed to 600 MPa. At a stress of 440 MPa, a significantly longer lifetime was observed in samples that had not been pre-stressed than was observed in pre-stressed samples. At a stress of 500 MPa, far longer lives were observed for some of the samples, which had not been pre-stressed, than was found in any of the pre-stressed samples. On the basis of these observations it appears that pre-stressing does reduce the fatigue life of samples, if they are fatigued above 380MPa, but because of the large scatter it is not possible to quantify the effect.

Table showing the maximum fatigue lives of samples of Ti44Al8Nb1B, which have been tested in fatigue either after pre-stressing to 600 MPa or without any pre-stressing.

σ_{max}, MPa	Pre-stressed	Un-pre-stressed
500	59,000	775,600
440	413,000	(ii) 6,020,000 +tinted+184,000
380	(i)tinted + 10,000,000+19,700@500MPa	Not tested
	12,000,000	Not tested

The tinting test labelled (i) in the Table was done after pre-stressing to 600 MPa and the sample was then fatigue tested at σ_{max} =380 MPa. After run out (> $1x10^7$ cycles) the test was interrupted and the sample was failed by cycling at 500MPa. The fracture surface clearly shows the tinted cracks which were initiated by pre-stressing. It is clear therefore that these cracks (which are arrowed in Fig. 4) have initiated the fatigue failure. Fig.4b shows the precracked area at a higher magnification. It can be seen that this pre-crack has a size of ~150 µm, which encloses four lamellar grains (the average grain size is about 75µm) and these grains are oriented such that the lamellae are parallel to each other.

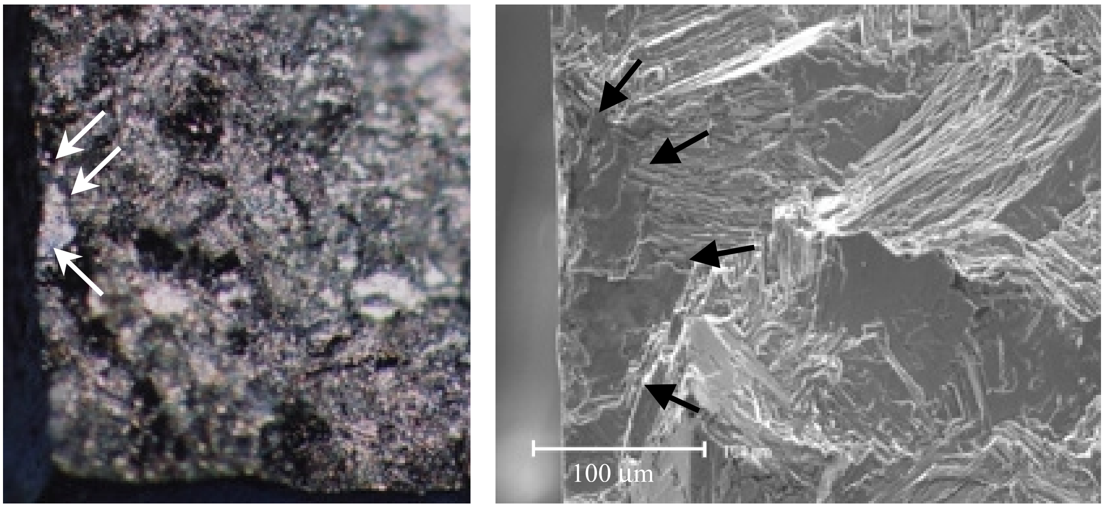


(a) (b)

Fig 4 (a) Optical and (b) secondary electron scanning electron micrograph obtained from a sample of Ti44Al8Nb1B which was prestressed at 600MPa, tinted, fatigued at 380MPa for > 10^7 cycles and then tested to failure at 500MPa. Note the large area defined by the crack which encompasses several grains where the lamellae are closely parallel.

In addition, a fatigue test at 440MPa on an un pre-stressed sample, which had been cycled at $6x10^6$ cycles, was interrupted and tinted at 700 °C/0.5h, as labelled (ii) in the Table. This sample was then cycled at 440 MPa until it failed after a further 184,000 cycles. The fracture surface is shown in Figs 5(a) and (b) show that a crack, with a depth of < 75 µm and a length of 150 µm on the surface, has been initiated after cycling at 440 MPa. This crack consists of remnants of two lamellar grains and other parts of these two grains have been machined off. It is also noted that the surrounding grains are all oriented at angles approaching 90° with respect to the grains in which the crack initiated, i.e., the tinted grains. This sample had a significantly longer lifetime than the others, which were tested under the same condition.

These tinting experiments show conclusively that cracks introduced during pre-stressing at 600MPa initiate the fatigue failure at 500 or 440MPa.The observations also suggest that the large scatter associated with the samples tested in fatigue is due to the fact that the cracks, introduced at the surface can be of varying length (depending upon the size of the (remnant) grain in which the crack is initiated. In addition the crack growth rate will be controlled by the angle that the lamellae in the adjacent grains make with the lamellae in the cracked grain.



(a) (b)

Fig 5 (a) Optical and (b) secondary scanning electron micrograph obtained from a sample of ti44Al8Nb1B which was fatigued at 440MPa for 6 x 106 cycles, tinted and then tested to failure at 440MPa. Note the small size of the initial crack and the large angle between the different parts of the crack as it grows through grains where the lamellae are at large angles to each other.

The formation of cracks at relatively low stresses raises several obvious questions concerning the lifing of these alloys for critical applications. It is noted that the ratio of fatigue limit to the proof stress in Ti44Al8Nb1B is only 0.65, whereas in a weak TiAl alloy, such as Ti48Al2Mn2NbB, where $\sigma_{0.2}$=350 MPa, the ratio can reach more than 0.9 for a similar microstructure. The fact that the fatigue limit is high in the weaker alloys shows that fracture does not occur and/or that any cracks formed do not propagate, despite the fact the dislocations are activated in the soft grains. It is therefore suggested that the actual stress which is required to cause fracture is approximately the same in the weak and the strong TiAl alloys and that this stress is not reached in the weaker alloys for the conditions under discussion.

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