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# NUCLEI FOR DUCTILE FRACTURE IN TITANIUM

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## INTRODUCTION

There have been a number of reviews in recent years on the subject of ductile fracture [1 - 3], and one of the conclusions from these has been that the nuclei for ductile fracture, in nominally single-phase materials, are inclusions or precipitate particles. There is no convincing evidence for any other type of nucleus [3], e.g., vacancy clusters.

Titanium alloys would seem to present an imposing challenge to this viewpoint. Titanium is able to dissolve substantial quantities of most of the common impurity elements, so that a significant inclusion population is rarely present. By inclusions in this context is meant "macroinclusions", particles $\geqslant 1$ µm diameter, which are the primary agents in control of ductility and toughness [2 - 5]. Furthermore, replicated fracture surfaces of Ti alloys only occasionally reveal particles within ductile dimples [6]. In two-phase Ti alloys, there is evidence that ductile fracture can be nucleated at interphase interfaces [7], but for single-phase alloys there is no indication in the literature of the nuclei for ductile fracture. This paper reports experiments which focus on this problem, and also presents a discussion of alternative nuclei.

## EXPERIMENTS

Hot-rolled Ti-Aℓ alloy sheet was available from a previous investigation [8]; the single-phase 6.1 wt. % Aℓ alloy, with 700 wt. ppm O, was chosen for primary emphasis, with supplementary experiments on specimens of 2.1 and 4.0 wt. % Aℓ. Tensile specimens with gage sections 1 x 5 x 40 mm were cut from the sheet with the tensile axis parallel to the rolling direction, and were annealed in 0.13 mPa vacuum for 8 hours at 50 K below the $\alpha/\alpha + \beta$ transus, producing a single-phase grain size of 35 µm. Specimens were tested at 77 K, 203 K and 298 K; ductility increased somewhat with temperature, but all fractures were visually similar. Fracture surfaces were examined in a scanning electron microscope (SEM), and in all cases the viewing direction was adjusted to be normal to the macroscopic fracture surface.

## RESULTS

Microscopically, ductile fracture was observed in all cases, although the detailed fracture morphology differed. For Ti-6Aℓ at 77 K, dimpled areas were predominant, with a few smooth featureless areas, Figure 1a. At 298 K,

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the fracture surface had, in addition to dimpled regions, a high density of tear ridges [9] oriented perpendicular to the flat face of the tensile specimens, with increased numbers of smooth areas, Figure 1b. Dimples at 298 K tended to be elongated parallel to the width axis of the specimen; orientation of these was opposite on opposite faces of the fracture, indicating a shear component in the microvoid coalescence process [9]. Fracture appearance at 203 K was intermediate between Figures 1a and 1b.

At higher magnifications, it was found that very few dimples contained evidence of a nucleating particle. Detailed dimple morphology varied with temperature. At 77 K, Figure 2 shows the relatively smooth and shallow character of the dimples. At 203 K, dimples were deeper and internally rougher, Figure 3, while at 298 K, they were also elongated, again indicating their partial shear character (Figure 4). The absence of particles here, it should be noted, constitutes no particular evidence on whether the dimple nuclei were in fact particles. It is frequently observed that particles are lost from fracture surfaces even when there is compelling evidence for particle nucleation [2, 3, 10, 11].

In considering various potential nuclei for microvoids, the dependence of fracture morphology on temperature plays an important role [3, 12]. Figures 1 - 4 clearly depict a decreasing proportion of shear fracture with decreasing temperature; they also show increasingly better-defined microvoid coalescence as temperature decreases. These are precisely the features described by Chin et al [12] for ductile fracture in high-purity aluminum containing no macroinclusions and very few microinclusions. Similar observations have been made by Miller and Besag [13] on very high purity aluminum and lead. These results for Ti-6Aℓ therefore suggest (i) that there is a strong possibility of microvoid nucleation by microinclusions, and (ii) that vacancy condensation to form nuclei is quite unlikely [3, 12, 13].

Specimens of Ti-2Aℓ and Ti-4Aℓ were also fractured at 77 K. Fractures were generally similar to the Ti-6Aℓ samples, as shown in Figures 5 and 6, although tear ridges were more evident and dimples were somewhat deeper and more varied in size. As a final point of comparison to the Ti-6Aℓ, a two-phase specimen of Ti-6Aℓ-4V (solution treated and aged) was examined after fracture at 298 K. As Figure 7 shows, fracture in this case is far more complex than in Figure 1b, and defies easy characterization. This result serves as a caution that results for single-phase Ti alloys cannot be extended to two-phase alloys.

## DISCUSSION

There are two points raised by this work for which discussion is appropriate. These are: the reasonableness of microvoid nucleation by small inclusions in a Ti alloy, and the reasonableness of the alternatives.

There is general agreement among investigators who have examined Ti alloys with transmission electron microscopy that inclusions are very infrequently observed; the suggestion of particle-nucleated voids is largely based on analogy with other materials. However, the absence of observations of particles is not as serious as it may seem at first, since the mean distance between 100 Å particles [3], located on a cubic array 5 μm on a side (the dimple diameter of Figure 2), is about 45 μm in a 2000 Å-thick foil. This leads to a sizeable sampling statistics problem in thin foils, which may account for the lack of observations of particles. From a phenomeno-





logical standpoint, the increased tendency to form microvoids at 77 K compared to 298 K is consistent with particle nucleation, since the strength of the matrix increases more rapidly with decreasing temperature than does the strength of the particle-matrix interface [12]; thus the stress required to separate a particle from the matrix is more readily achieved at lower temperatures or higher strength levels. Consistent with this interpretation is the increasing shear character of voids formed with increasing temperature. Thus, while the evidence for particle nucleated voids is indirect, the suggestion is consistent with the present fracture observations.

In view of the absence of direct evidence for particle nucleation of voids, alternate types of void nuclei should be considered. Several possibilities exist, including twin-matrix interfaces, vacancy clusters, grain boundaries (and grain boundary-slip band intersections), and slip band - slip band intersections. If twin-matrix interfaces are void nuclei, then the void size and distribution might be expected to depend on twinning frequency and on twin volume fraction. This is not the case, as can be seen by comparing the Ti-2Aℓ and Ti-6Aℓ fracture characteristics; there is a large decrease in the extent of twinning when the Aℓ content is increased from 2% to 6% [8]. The role vacancies can play in nucleating voids has been discussed elsewhere and must be considered unlikely in the present tests [3, 12, 13]. Grain boundaries also appear unlikely here, since the fracture surface is not observed to follow grain boundary paths. Slip bands thus appear to be the only real alternative to particles as void nuclei in these Ti alloys.

It is difficult to assess slip band intersection in detail (we refer here to microscopic slip bands, not to macrobands [14] whose behaviour may be controlled by particles), and we discuss it here as a possibility only. It is very significant, however, that the hexagonal lattice has a very limited capacity for relaxing stress concentrations at intersecting slip bnads by means of plastic flow with a component parallel to the axis, so that local stresses can be high. These stresses could lead to dislocation coalescence of the kind envisaged by Stroh [15], which in turn could provide void nuclei. Increasing Aℓ content should then cause shallower dimples (cf. Figures 2, 5, and 6). However, void sheet formation [3, 10, 11] in these clean materials would be controlled by the same factors, so this observation is not decisive.

Finally, it is appropriate to comment on the relationship between void nucleation and growth in single-phase Ti alloys and the ductile fracture observed in the more technologically important $\alpha + \beta$ Ti alloys. While void nucleation at particles or slip band intersections could operate in $\alpha + \beta$ alloys, there is reason to believe that this is overshadowed by void nucleation at interphase interfaces. Such interfaces comprise not only the boundaries between primary $\alpha$ and the transformed $\beta$ matrix [7], but also Widmanstätten $\alpha$ plate boundaries, especially in high-strength alloys [16]. Thus, as Figure 7 illustrates, the present observations on single-phase alloys should not be extrapolated to two-phase Ti alloys.

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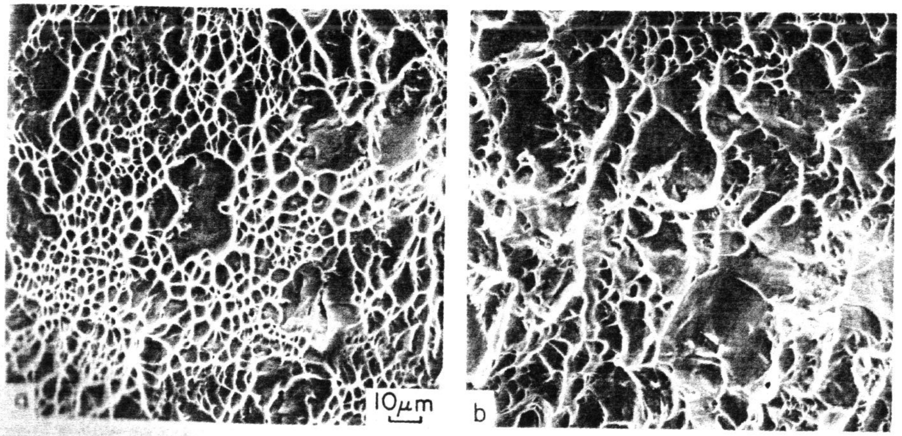


Figure 1 SEM Views of Fracture Surface of Ti-6Aℓ, Same Magnification (a) 77 K Fracture, (b) 298 K Fracture





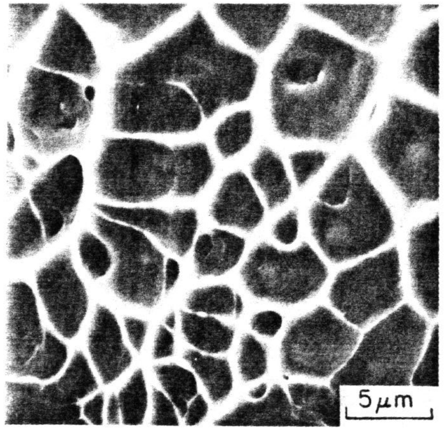


Figure 2 SEM Fractography of Ti-6Aℓ Fractured at 77 K, Illustrating Dimple Character

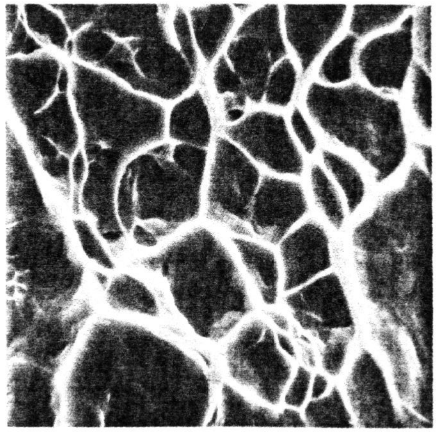

Figure 3 SEM Fractography of Ti-6Aℓ Fractured at 203 K; Same Magnification as Figure 2

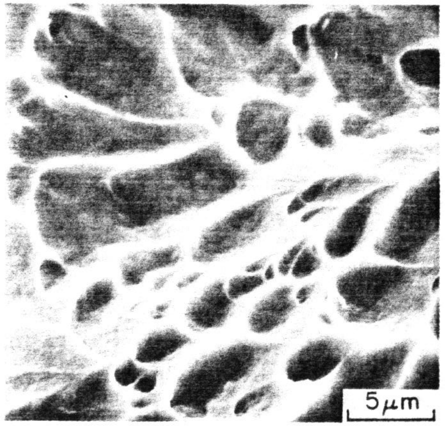


Figure 4 SEM View of Dimples in Ti-6Aℓ Fractured at 298 K

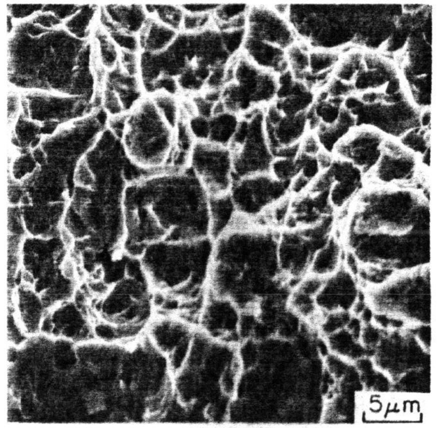


Figure 5 SEM Observation of Ti-2Aℓ Fractured at 77 K

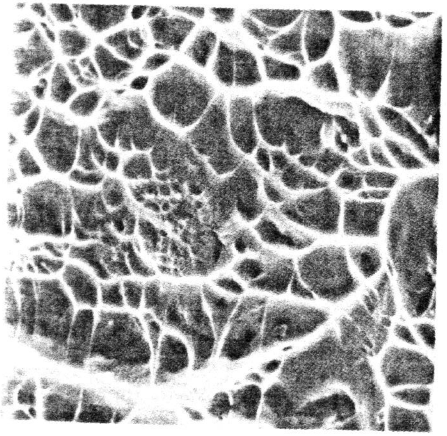

Figure 6 SEM Observation of Ti-4Aℓ Fractured at 77 K; Same Magnification as Figure 5

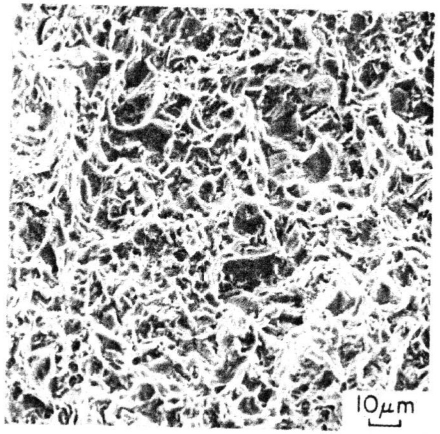


Figure 7 SEM Fractography of Solution Treated and Aged (STA) Specimen of Ti-6Aℓ-4V, Fractured at 298 K Under Plane Strain Conditions