PROCESSING AND PROPERTIES OF DUCTILE PHASE
REINFORCED GAMMA TITANIUM ALUMINIDE ALLOYS

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Abstract

Gamma Ti aluminide (γ, TiAl) alloys offer significant potential for a variety of aerospace applications, however, their use is limited in part by low ductility and fracture toughness. A powder metallurgy (P/M) composite approach, involving the blending of a γ alloy with a ductile/tough Ti-Nb alloy powder has shown potential for increasing the extrinsic fracture toughness of γ. In this study, moiré interferometry was used to measure fracture toughness behavior for monolithic and phase blended γ alloys. This whole-field measurement technique was used to record both vertical and horizontal displacements associated with crack growth and to evaluate the crack opening displacement (COD) for compact tension specimens. The results showed that the fracture toughness of phase blended γ sheet was noticeably increased, while ductility remained similar to that of as-HIP’ed monolithic γ. The increase in fracture toughness in the phase blended γ is due to formation of crack face bridging, crack path deviation, blunting and multiple crack formation.
I. Introduction

Phase blending has been proposed as a method of improving the fracture toughness of Ti aluminides [1]. Phase blending involves incorporating a ductile/tough phase into the matrix of a brittle material to serve as a crack arrestor or to restrict crack propagation by forming ductile ligaments to bridge the crack front [2]. Traction produced by the bridges reduce crack tip stress intensities. Ductile phase reinforced composites often experience large-scale bridging effects [3], especially for short cracks or in components with small dimensions. The overall load displacement characteristic \( P(\delta) \) of a reinforced component is governed by traction law \( p(u) \) for the ductile phase, as well as the effective crack tip fracture resistance of the composite, \( \Gamma_t \) and the elastic modulus \( E' \) [4]. In addition, important secondary toughening mechanisms have been demonstrated, which arise from crack particle interaction processes [4]. These mechanisms include crack trapping and renucleation leading to increases in \( \Gamma_t \) and crack branching and deflection. The toughening contributions from these mechanisms, depend upon the combined properties of the ductile phase, interface reaction layer and the matrix. Numerous studies of the fracture behavior of brittle matrices containing ductile particles show that, if strongly bonded to the matrix, the ductile particles act to toughen the matrix. The degree of toughening increases with volume fraction and ductile particle size. Difficulty in applying phase blending to \( \gamma \)-TiAl alloys has stemmed from excessive interdiffusion and reaction of the constituents, during both consolidation and mechanical working at elevated temperature. Typically, the high temperatures required to consolidate, extrude, or roll phase blended \( \gamma \)-TiAl alloys results in a variable composition within the microstructure (i.e., between the matrix and the toughening constituent) and reaction product formation at the \( \gamma \)-TiAl/ductile phase interface. In addition, deformation of the powder mixture produces a nonuniform distribution and morphology of the ductile phase particles. These factors limit the effects of phase blending. If, however, these limitations can be reduced through proper selection of the constituent chemistries and thermomechanical processing route, the resulting \( \gamma \) alloy, having an improved balance of engineering properties, would prove very attractive for light weight, high temperature structural applications. It is the primary purpose of this study to explore microstructural development and crack propagation behavior in a \( \gamma \) system, using a moiré interferometry, a whole-field strain mapping technique, in concert with a detailed microstructural analysis.

A collaborative effort at Rockwell International Science Center (RISC) and Nuclear Metals, Inc. (NMI) has been carried out to explore powder metallurgy techniques for the fabrication of "phase blended" \( \gamma \)-TiAl sheet. The NMI PREP™ powders exhibit excellent compositional uniformity with low interstitial content, and may be useful in producing a sound composite with good fracture toughness. In this study, the microstructures and properties of monolithic \( \gamma \) and a mixture of 80 volume% \( \gamma \) + 20 volume% Ti-Nb powder were HIP consolidated at 1000°C/300 MPa for characterization. This low temperature/high pressure condition was selected to minimize interdiffusion and extensive reaction at the \( \gamma \)/Ti-Nb interface.

II. Experimental

Materials and Processing

The \( \gamma \) alloy selected for this study was a two-phase, \( \gamma + \alpha_2 \), alloy of composition Ti-48Al-2.5Nb-0.3Ta (at.%). The toughening powder constituent was an alloy of composition Ti-30at.%Nb, having primarily a \( \beta \)-Ti microstructure with a minor volume fraction of \( \alpha \) at low temperature. These powders were produced by NMI using a plasma rotating electrode (PREP™) process. The powders were sieved prior to blending, with -250 \( \mu \)m (-60 mesh) \( \gamma \) powder and 180-500 \( \mu \)m (+80/-35 mesh) Ti-Nb powder being used for the study. Both monolithic \( \gamma \) and a mixture of 80 volume% \( \gamma \) + 20 volume% Ti-Nb powder were HIP consolidated at 1000°C/300 MPa for characterization. This low temperature/high pressure condition was selected to minimize interdiffusion and extensive reaction at the \( \gamma \)/Ti-Nb interface.

Tensile Testing

Machining of the specimens was performed at MET-CUT Research Associates in Cincinnati, Ohio. Flat tensile specimens, having a nominal gauge section of 2 mm x 9.4 mm x 25.4 mm
long, were tested at room temperature. Tensile test pieces were machined using low stress grinding techniques. Two tensile specimens were tested for each material. Tests were conducted at room temperature on a 20 kip closed loop servo-hydraulic test machine per ASTM E8. Strain control was set so that a strain rate of 0.005 / sec was maintained through 0.2% yield and the head rate was 0.05 in./min. after that. Extensometers were used on all specimens.

**Crack Growth Experiments (Resistance Curves and Fracture Toughness)**

Cracks were grown stably under measured loading conditions in compact tension specimens of dimension 30 x 32 x 2 mm. A loading fixture that could attach to the stage of an optical microscope was used, to allow in situ observation of the crack tip and wake zones. A four-beam moiré interferometry system was used to measure two orthogonal in-plane displacement fields simultaneously [5]. This technique involves replicating a diffraction grid to the surface of the compact tension specimen before growing the crack, and then, during loading to extend the crack, illuminating the surface with an Argon-ion laser (wavelength 514 nm). In this set up, part of the incident beam impinged directly onto the specimen surface, while the other part, after reflection from three plane mirrors, was incident in a symmetrical direction to produce a virtual reference grating. The combination of this reference grating and the diffraction grating bonded to the specimen produced the moiré pattern, which was recorded photographically. The moiré pattern consists of a set of fringes, which represent contours of constant displacement in the x1 and x2 directions, the increment between adjacent fringes being equal to the period of the reference grating (0.417 µm). In the present experiments, this technique allowed strain resolution of 10⁻³ over gauge lengths as small as 40 µm. The testing environment for toughness testing was ambient air. Digitized recordings of the moiré patterns were converted to strains automatically [6]. The evaluation of the J-Integral is essentially a numerical integration along a loop encompassing the crack where the three strain components must be evaluated at identical points along the chosen path [6]. J-Integral requires the strain components, the stress components and the strain energy density. The three stress components are calculated using J₂-deformation theory of plasticity for multi-axial states with a power hardening stress-strain relation. Stress intensity factors were evaluated using k=(JxE)₀.⁵. Fracture toughness values were obtained by marking the maximum in Kₐ vs. crack extension curves.

**III. Results**

The HIP-consolidated γ alloy formed a fine grained (~20 µm) equiaxed microstructure, while the γ + Ti-Nb phase blended microstructure comprised spherical γ-TiAl regions (of approximately the same 20 µm size) encompassed by a nearly continuous β-Ti phase. This is most easily seen in the back-scattered electron image of Figure 1a, in which the Ti-Nb region is lighter in contrast with the darker γ particles. In addition, an approximately 10 µm thick interdiffusion/reaction zone formed between the phase blended constituents, as shown in Figure 1b. The monolithic γ and phase blended γ + Ti-Nb sheets had measured densities of 3.76 and 3.97 g/cc, respectively.

The uniaxial stress-strain relations for monolithic and phase blended alloys with the two coefficients for the power hardening relations are shown in Figure 2a and 2b, respectively. Room temperature elongations were below 1% for both the monolithic and phase blended sheet. All fracture surfaces showed occasional flat cleavage facets and ductile tearing. There was no evidence of intergranular fracture. Although the phase blended fracture surface was more rough, grain boundary facets and prior powder boundary facets were not observed.

Crack opening profiles for monolithic and composite γ-TiAl alloys, measured by moiré interferometry, are shown in Figures 3a and 3b, respectively. Comparison of the profiles show that in the composite material the crack is bridged. In this case bridging is supplied by ductile ligaments that remain connected to both fracture surfaces behind the crack front. The bridging traction could be determined from measurements of such crack opening profiles [7]. Figure 4 shows crack mouth opening displacement (CMOD) vs. crack length. The phase blended alloy shows larger CMOD, because it has lower elastic modulus and higher ductility than the monolithic alloy.
$K_R$--$\Delta a$ resistance curves are shown in Figure 5. For the monolithic alloy crack extension has practically no effect on the toughness, when the crack is fully grown out of the semi-chevron zone. The average toughness for monolithic alloy is $=11 \text{ MPa} \cdot \text{m}$. In contrast, some resistance curve behavior is observed for the phase blended composite. $K_R$ increases from an apparent initiation value of $=15 \text{ MPa} \cdot \text{m}$ to $=19 \text{ MPa} \cdot \text{m}$ at $\Delta a \geq 10 \text{ mm}$.

Figures 6a and 6b show micrographs taken along the crack path in the compact tension specimen. Ductile ligaments along the fracture surface remain intact as shown in Fig. 6a. These ligaments simply pull out when the bridging zone is saturated. With reference to work of Ashby [8], by allowing for the constraint of the surrounding matrix, the average flow stress of the ductile particles increases significantly. The crack growth sequence in the phase blended alloy is: crack grows in $\gamma$ phase toward the reaction zone, crack blunts when it grows into ductile zone, crack re-nucleates behind the ductile ligament, (at this point crack is bridged), and series of microcracks develop parallel to the second crack and grow toward Ti-30Nb phase (Fig. 6b). Blunting and microcrack development allow the particles to stretch out as the crack faces separate, forming bridging ligaments that hinder the crack advance. Due to these ligaments, the value of $K_{IC}$ will increase as the crack length increases up to the point at which the Ti-30Nb particle pulls out by separation along the reaction zone.

IV. Discussion

Bridging is the most puissant way of improving the toughness of brittle materials. The physical mechanism of toughening is straightforward: if ductile particles span the advancing crack, they must stretch as the crack opens until they fracture or decohere: the work-of-stretching contributes to the overall toughness of the solid. If the particle is so weakly bonded to the matrix that it easily pulls free as the crack approaches, then it is not stretched and there is almost no contribution to the toughness. But if it is strongly bonded, it is constrained. Its force-displacement curve is then very different from that of the unconstrained material as measured in an ordinary tensile test. This is an important difference because the energy absorbed in stretching the particle, crucial in calculating the contribution to the toughness, depends strongly on the degree of constraint. Based upon the microstructural observations, the reaction zone boundaries in the phase blended sheet are the weakest link between the two constituents. Therefore, optimization of the HIP cycle for minimization of interdiffusion between the $\gamma$ and Ti-30Nb constituents is crucial. In addition, altering the composition of the matrix and/or the toughening constituent may modify the kinetics of interdiffusion/reaction and thereby modify the interface strength characteristics.

The sheet ductility's were all below 1% demonstrating the need for subsequent thermomechanical processing, as has been found in other studies on compaction of Ti aluminide powders. Thus as-HIP'ed ductility of the matrix appears to be a controlling factor in the overall behavior of this material. The yield stress of the phase blended sheet was slightly lower than the monolithic sheet, but the ductility was somewhat improved. The nearly continuous ductile Ti-30Nb phase would account for this. The toughness of phase blended sheet was greatly improved demonstrating the effectiveness of the Ti-30Nb phase. The quantitative contribution of different mechanisms of fracture toughness enhancement is now being investigated.

V. Conclusions

It has been demonstrated that PREP™ powders of $\gamma$ and a ductile binary Ti-30Nb alloy can be physically blended and compacted into thin sections using hot-isostatic-pressing.

Compared to monolithic $\gamma$ sheet, the phase blended sheet showed a slight increase in ductility and a noticeable increase in fracture toughness. Techniques to assure homogeneous distribution of the ductile phase and improve the as-HIP'ed ductility must be developed to realize the full potential of this material.

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VII. References


Figure 1. Micrographs of the as-HIP'ed phase blended γ + Ti-30Nb alloy showing the morphology and distribution of the toughening constituent (1a) and the interface microstructure development (1b).
Figure 2. The uniaxial stress-strain relations for a) monolithic and b) phase blended alloys.
Figure 3. COD profiles for a) monolithic, and b) composite γ-TiAl alloys, measured by moiré interferometry.
**Figure 4.** Crack mouth opening displacement (CMOD) vs. crack length.

**Figure 5.** $K_R$-$\Delta a$ curves for Monolithic and Phase blended alloys.
Figure 6. Micrographs of phase blended alloy taken along the crack path.