NEW γ-TiAl ALLOYS CONTAINING Ni

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Introduction

The solubility of single phase γ TiAl whose stoichiometric composition is Ti₃₂Al₃₈ is expanded into the Al rich side of a Ti-Al phase diagram [1], i.e., the range is approximately 49 - 56 at % Al at 500°C. The Ti rich compositions, the other side of the stoichiometry, are stable in a mixture of two phases (α₃ + γ). The mechanical properties of Al rich γ TiAl alloys were characterized as hard and brittle at room temperature and intermediate temperatures [2,3]. The elongation of polycrystalline TiAl at room temperature is as small as ~0.1% [2]. In the Al rich single phase TiAl alloys, the hardness increases [4-6] and the fracture strain decreases [5,6] with increasing Al content. It appears that the directional bonds (Ti-Al) [7] may be enhanced with the excess Al and thereby, decrease the mobility of dislocations in these alloys. If such a role of Al in the mechanical properties can be assumed for hypothetical single γ phase Ti rich Ti-Al compositions, an improved fracture strain would be anticipated due to the increased metallic bonds.

On the other hand, polysynthetically grown Ti rich Ti-49.3 at %Al alloy by an optical floating zone technique exhibits an orientation dependent deformation [8], where the twinning plays an important role in deformation. Conventionally processed Ti rich binary Ti-Al alloys or multicomponent two phase titanium aluminides whose microstructure was identified with lamellar structure and γ phase grains exhibit moderate ductility i.e., tensile elongation 1-3.5% [9]. A two phase Ti₂₃Al₆₈ alloy which was processed by rapid solidification (RS) and consolidated exhibits a bend ductility of 0.9% (four point bend tests) [10]. For this alloy, no significant effect of rapid solidification on ductility was observed. For this alloy, no significant ductility (above 4% tensile elongation) was obtained with the lamellar structure in the Ti rich Ti-Al alloys regardless of the processing methods employed.

A recent study [11,12] shows that there exists a significant difference in dislocation structure between Al rich γ-TiAl alloys and Ti rich γ phase grains. For instance, the decomposition of superdislocations occurs at room temperature in Ti rich γ phase grains whereas the same reaction only occurs at high temperatures in Al rich γ TiAl. This evidence suggests that the dislocation reaction is more active in Ti rich single γ phase grains than in Al rich γ TiAl alloy matrix at room temperature. The authors have investigated single γ phase titanium aluminides through rapid solidification regarding the solubility of alloying elements [13] and mechanical properties. Some of the results of Al lean Ti-Al alloys on the mechanical properties are presented in this article.

Experimental

Small buttons of TiAl alloys containing Ni or Ni, Cr, Nb and V, with Al less than 50 at % were prepared in an arc furnace by repeated melting. The alloy buttons were broken into small pieces (10-30 mg) and were
then splat-quenched into thin foils of 20-30 \( \mu \text{m} \) thick in an arc furnace. The foils were encapsulated in a quartz tube in vacuum and annealed at 1000\( ^\circ \)C for one day.

Bulge tests, which are similar to "punch-stretch", tests were performed on thin disk-shape foils whose rims were tightly held between two ring-shaped specimen holders while the center of the foil specimen was allowed to deform by a small ball to failure (Fig. 1a & b). The strain at the failure point is defined as the fracture strain. The tests were performed on an average of ten specimens to determine each data point of the fracture strain. The fracture strain (percent) was calculated from the following geometrical relation:

\[
\varepsilon(\%) = \frac{h^2 + h t}{a^2 + h^2 + h t} \times 100
\]

where the dimensions \( h, t \) and \( a \) are indicated in Fig. 1b.

The bend ductility was determined using a metal cone which provided various radii. The specimen of 2-3 mm wide and 10-15 mm long was pressed against the convex surface of the cone to introduce bend strain to the specimen. The specimen continued to move from a larger circumference to a smaller one until the failure occurred. The bend strain was calculated from the thickness and the radius of the circumference at which the failure occurred.

**Results**

Optical micrographs (Fig 2a & b) show a uniform grain size of in the annealed Ti\(_{51.8}\)Al\(_{47.7}\)Ni\(_{0.5}\) alloy foil. The grain size is \( \sim 12 \mu \text{m} \) in its cross section (a), and \( \sim 18 \mu \text{m} \) in the flat surface (b), respectively. No significant grain alignment in its cross section was observed. Also, fine, unknown precipitates were dispersed in the matrix. No lamellar structure was observed in any grains. The detailed microstructure studied by TEM is shown in Fig. 3 in which the single \( \gamma \) phase matrix is decorated with random precipitates at the annealed state (Fig. 3a).

The annealed foils were deformed by compression and were studied by TEM. The microstructure shown in Fig. 3b shows well developed dislocation structures. In particular, dislocation networks labelled A and B shown in Fig. 3b were identified as having a very fine scale i.e., the distance between the dislocations is on the order of 10 nm. Such a fine scale dislocation network was not observed in Al rich \( \gamma \) titanium aluminides.

The bulge tests and bend tests performed on various Ti-Al alloy foils are tabulated in Tables 1-3. The fracture strain in these tables consists of both elastic and plastic deformation, but the elastic portion of the figures may be small (typically 0.1-0.2 %). The fracture strain of the ternary alloy in Table 1 increases with Ni content and decreases again beyond 0.5 at % Ni. However, in the V containing multicomponent alloy in Table 1, the maximum strain occurs at 1.0 at % Ni. In Table 2, the maximum bend ductility of the ternary alloy occurs at 2.0 at % Ni. This as-quenched alloy exhibits a complex x-ray diffraction pattern which was different from those measured for single \( \gamma \) phase material. It is probable that the alloy in the Table 2 consists of more than one phase. Finally, in Table 3, the hardness and fracture strain of the RS alloy and the conventional multicomponent alloy were compared with respect to hardness and fracture strain.

### Table 1: Bulge Tests

<table>
<thead>
<tr>
<th>RS (Ti(<em>{52})Al(</em>{48}))(<em>{100-x})Ni(</em>{x})</th>
<th>( x=0.0 )</th>
<th>0.25</th>
<th>0.5</th>
<th>1.0</th>
<th>2.0</th>
</tr>
</thead>
<tbody>
<tr>
<td>As-quenched state (%)</td>
<td>1.97</td>
<td>2.3</td>
<td>5.5</td>
<td>3.0</td>
<td>2.3</td>
</tr>
<tr>
<td>Annealed state (%)</td>
<td>2.96</td>
<td>5.5</td>
<td>8.5</td>
<td>4.5</td>
<td>3.1</td>
</tr>
</tbody>
</table>

<table>
<thead>
<tr>
<th>RS Ti(<em>{50})Al(</em>{47})V(<em>{10})Nb(</em>{0.5})Cr(<em>{1.5})Ni(</em>{x})</th>
<th>( x=0.5 )</th>
<th>1.0</th>
<th>2.0</th>
</tr>
</thead>
<tbody>
<tr>
<td>As-quenched state (%)</td>
<td>1.15</td>
<td>1.27</td>
<td>0.6</td>
</tr>
<tr>
<td>Annealed state (%)</td>
<td>2.1</td>
<td>4.6</td>
<td>1.6</td>
</tr>
</tbody>
</table>
Table 2: Bend Ductility

<table>
<thead>
<tr>
<th>Thickness Reduction (%)</th>
<th>x=0.0</th>
<th>1.0</th>
<th>2.0</th>
<th>4.0</th>
</tr>
</thead>
<tbody>
<tr>
<td>RS (Ti,8Al,4)Ni, (As-Quenched)</td>
<td>2.5</td>
<td>2.6</td>
<td>4.7</td>
<td>1.6</td>
</tr>
</tbody>
</table>

*: The ductility was determined on the as-quenched specimens.

Table 3: Comparison Between Conventional Alloy and New Alloy

<table>
<thead>
<tr>
<th>Hardness (Gpa)</th>
<th>Bulge strain %</th>
<th>Bend strain %</th>
</tr>
</thead>
<tbody>
<tr>
<td>RS (Ti,8Al,4)Ni, (Annealed)</td>
<td>3.6</td>
<td>8.5</td>
</tr>
<tr>
<td>Ti-Al,4Cr,7V,9Nb,3 (Conventional processing)</td>
<td>2.75</td>
<td>3.2*</td>
</tr>
<tr>
<td>RS Ti,8Al,4 (Consolidated)</td>
<td></td>
<td></td>
</tr>
</tbody>
</table>

*: the same bulge tests (Fig. 1) were performed on thinned specimens (50-70μm thick).
**: Four point bend tests.

Discussion

The fact that as-quenched (Ti,8Al,4)Ni, is a γ phase with a small amount of precipitates indicates that the solidification path includes a single γ phase region. The γ phase boundary apparently was shifted toward Ti rich compositions with a small amount of Ni. However, it is not known whether an eutectoid transformation exists between 1000°C and room temperature.

Addition of a small amount of Ni to Ti,8Al,4 alloy has a large influence on both the microstructure and mechanical properties of the alloy. Specifically, the lamellar structure and γ grains in Ti,8Al,4 Alloy are replaced by a single γ phase matrix decorated with a small volume fraction of precipitates. The single γ phase grains are able to produce high density dislocations (Fig. 3b), which may be responsible for the high fracture strain. The empirical findings show that addition of Ni on ductility is effective in those titanium aluminides where the amount of Al is less than 50 at% and where the combined amount of Ti and early transition elements is more than 50 at%. On the other hand, the high hardness of (Ti,8Al,4)Ni, alloy may be attributed to the fine precipitates observed in the otherwise single γ phase matrix as shown in Fig. 2a. Both hardness and fracture strain are higher in the RS alloy with Ni than in the conventionally processed multicomponent alloy (Table 3). The differences between these latter two alloys arise from the fundamentally different microstructures, i.e., lamellar structure vs. single γ phase structure with Al less than 50 at%.

Conclusions

1) The addition of a small amount of Ni (0.5 at%) to the Ti,8Al,4 alloy results in a largely single γ phase structure at 1000°C.

2) The largely single γ phase Ti,11.6Al,6.8Ni,3 alloy which was rapidly quenched and annealed at 1000°C for one day was deformed to 1-2% under compression. In this alloy, high density dislocation networks were formed in the γ phase matrix. Low density dislocation networks are observed in Al rich γ TiAl alloys.

3) Stacking faults and twins were rarely observed in the Ti,11.6Al,6.8Ni,3 alloy, indicating that the role of these planar defects is small during deformation. The shape accommodation during deformation of this alloy comes primarily from dislocation creation and motion.

4) Maximum fracture strains of 8.5% and 4.2% occurred at x=0.5 in (Ti,8Al,4)Ni, and at x=1.0 in Ti,8Al,4Ni, respectively. In general, the alloys which contained Ni studied...
here exhibited improved ductility either in the as-quenched state or in both the as-quenched and annealed states.

References


Fig. 1 A small bulge tester for foil specimens: a) sketch of the tester b) specimen geometry during testing.

Fig. 2 Optical micrographs of (Ti52Al48)99.5Ni0.5 alloy foil after heat treatment at 1000°C for one day; a) the cross-section of the specimen, b) a flat surface of the specimen.
Fig. 2b

Fig. 4 Plot of fracture strain vs. Ni content in the annealed (Ti_{52}Al_{48})_{100-x}Ni_{x} alloy.
Fig. 3 TEM micrographs of (Ti$_{1.2}$Al$_{0.8}$)$_{95}$Ni$_{5}$ alloy foil; a) after annealed at 1000°C for one day (B=[110], g=[220]); b) deformed grains in the specimen which was annealed and deformed to 1-2% (B=[011], g=[200]).