Weldability of a Titanium Aluminide

Control of the weld cooling rate was found to be a critical factor in optimizing the weld structure and properties

BY W. A. BAESLACK III, T. J. MASCORELLA AND T. J. KELLY

ABSTRACT. The weldability of an alpha-two titanium aluminide, Ti-13.5 wt-%Al-21.5 wt-%Nb (Ti-24 at.-%Al-11 at.-%Nb), has been investigated from a perspective of developing relationships between the weld cooling rate, microstructure, mechanical properties and fracture behavior. Dilatometry studies performed over a range of cooling rates from 1°C/s (1.8°F/s) to 150°C/s (270°F/s) showed a continuous decrease in the body-centered cubic (BCC) phase transformations. Water quenching from above the beta transformation start temperature, and cooling rates of 750°C/s (1350°F/s), which promoted complete retention of BCC phase.

Metallurgical characterization of the dilatometry specimens using light and transmission-electron microscopy (TEM) revealed beta decomposition to alpha-two platelets, which became progressively finer and more acicular with increasing cooling rate. At the slowest cooling rate of 1°C/s (1.8°F/s), the observation of both midribs at the centerline of alpha-two platelets and Nb-enriched strips of retained beta phase at alpha-two platelet boundaries suggested transformation by a mixed shear/diffusional mechanism. The apparent absence of retained beta phase in specimens cooled at rates from 5°C/s (9°F/s) to 150°C/s (270°F/s) indicated beta decomposition entirely by a diffusionless/shear transformation over this range of cooling rates.

Simulated gas tungsten arc (GTA) weld fusion and heat-affected zone (HAZ) specimens were generated using a Gleeble 1500 system by superimposing transformed-beta microstructures onto the prior-beta grain macrostructures of actual GTA weldments and beta-solutionized thin sheet. Microstructural characteristics of the simulated fusion and HAZ specimens produced at cooling rates from 5°C/s (9°F/s) to 50°C/s (90°F/s) compared well with those of the dilatometry specimens. Mechanical property testing found the hardness and tensile strength of the simulated weld structures to increase with increasing cooling rate and to exceed that of the alpha-two/beta-processed base metal for all cooling rates. Conversely, the generally low ductility levels of the weld zones as measured by three-point bend testing were found to decrease with increasing cooling rate, being optimum at the slowest cooling rate of 5°C/s. Fractographic analysis of fusion and HAZ structures produced at the lower cooling rates revealed a macroscopically flat appearance that microscopically exhibited both cleavage and ductile tearing. Increases in cooling rate promoted a macroscopically faceted fracture surface and an increase in the proportion of cleavage versus ductile tearing.

Implications of the continuous-cooling phase transformation study on the joining of Ti-13.5 wt-%Al-21.5 wt-%Nb using alternate welding processes are also discussed.

Introduction

The development of advanced, high-performance aircraft and aerospace systems, such as the transatmospheric National Aerospace Plane (NASP), will require the utilization of new, lightweight materials that can perform at temperatures well above those currently allowable with conventional elevated-temperature aluminum and titanium alloys (Ref. 1). During the past decade, the Air Force and aerospace industry have responded to this requirement with the development of a new class of aerospace materials that are based on the ordered Ti₃Al intermetallic compound, designated as "alpha-two." The Ti₃Al titanium aluminide has been shown to offer attractive strength, creep and elastic modulus properties at a density below that of conventional titanium alloys (Ref. 2). However, the inherent room-temperature brittleness of the binary Ti₃Al compound has precluded its utilization in engineering applications. Recently, ternary and quaternary alloying element additions and advanced processing techniques have been utilized to generate alpha-two titanium aluminide alloys that offer improved ductilities while retaining the attractive properties of the binary compound (Refs. 3, 4). Ti-13.5Al-21.5Nb (designated hereafter as Ti-14Al-21Nb) represents a first-generation alpha-two titanium aluminide that has experienced extensive evaluation and optimization from thermomechanical processing and heat treatment standpoints (Ref. 5).

Fusion welding is a cost-effective manufacturing method for many components in demanding aerospace applications. Over the years, numerous investigators have shown that by implementing a knowledge of welding metallurgy, high-integrity welds can be made in complex titanium-based alloys using a variety of arc and energy-beam processes. Despite the intended application of advanced alpha-two titanium aluminides in structural applications, relatively little work has addressed the fusion welding behavior and welding metallurgy of this alloy system. The purpose of the present study was to investigate the gas tungsten arc weldability of Ti-14Al-21Nb, as well as develop a basic understanding of the welding metallurgy of this new family of aerospace materials.
Background

Metallurgy of the Alpha-Two Titanium Aluminides

As the binary Ti-Al phase diagram in Fig. 1 shows, the Ti₃Al "alpha-two" intermetallic forms allotropically on cooling from the high-temperature body-centered cubic beta phase over a range of aluminum contents from about 22 to 36 at-% (Ref. 6). The compound exhibits a hexagonal close-packed crystal structure that is ordered (DO₁₉ superlattice). Compared to conventional elevated-temperature titanium alloys (Table 1), such as Ti-6Al-2Sn-4Zr-2Mo-0.1Si (Ref. 7), Ti₃Al exhibits an increased elastic modulus at room and elevated temperatures, improved creep properties and enhanced oxidation resistance. Unfortunately, the binary compound also exhibits an unacceptably low room-temperature ductility on the order of 1-2% elongation. This low ductility results principally from limited planar slip and an absence of twinning in the ordered structure (Ref. 4).

In recent years, extensive alloy development studies have sought to improve titanium-aluminide ductility through ternary and quarternary alloying additions (Ref. 3). These studies have determined that the addition of certain isomorphous beta-stabilizing elements, such as Nb, can markedly improve room-temperature ductility to levels required for engineering applications. Beta-stabilizer additions were found to increase ductility of the alpha-two phase by inhibiting ordering kinetics (anti-phase domain growth), reducing slip length and planarity and increasing the tendency for nonbasal slip (Ref. 4). In addition, the retention of a small quantity of beta phase in the microstructure further improves ductility by dispersing slip. Based on these alloy development studies, a first-generation titanium aluminide, Ti-14Al-21Nb, was developed. Figure 2 shows the Ti₃Al-Nb pseudo-binary phase diagram developed from the work of several investigators (Refs. 8-10). As indicated, the nominal Ti-14Al-21Nb chemical composition actually represents an alpha-two/beta alloy with a beta transus temperature of about 1135°C (2075°F) which differs from the 1000°C in Fig. 2 due to minor alloying element effects. In recent years, processing and application studies have shown that from thermomechanical processing and heat treatment standpoints, the alpha-two/beta microstructures generated in Ti-14Al-21Nb, via beta or alpha-two/beta processing morphologically resemble those exhibited by conventional alpha/beta titanium alloys of comparable beta stability (Ref. 5). General microstructure/mechanical property/fracture relationships have also been shown to parallel those associated with

**Table 1—Property Comparison for High-Temperature Alloys (Refs. 1, 4)**

<table>
<thead>
<tr>
<th>Property</th>
<th>Ti₃Al</th>
<th>Ti-14Al-21Nb</th>
<th>Ti-6Al-2Sn-4Zr-2Mo</th>
</tr>
</thead>
<tbody>
<tr>
<td>Density, g/cm³</td>
<td>4.15</td>
<td>4.6 (0.167)</td>
<td>4.5 (0.163)</td>
</tr>
<tr>
<td>(lb/in³)</td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>Young's Modulus, GN/m² (10⁶ psi)</td>
<td>145 (21.0)</td>
<td>98.5 (14.2)</td>
<td>114 (16.5)</td>
</tr>
<tr>
<td>Max Temp — Creep, °C (°F)</td>
<td>815 (1499)</td>
<td>600 (1112)</td>
<td>500 (932)</td>
</tr>
<tr>
<td>Max Temp — Oxidation, °C (°F)</td>
<td>650 (1202)</td>
<td>650 (1202)</td>
<td>500 (932)</td>
</tr>
<tr>
<td>Ductility — RT (%)</td>
<td>1-2(a)</td>
<td>2-5(a)</td>
<td>10(a)</td>
</tr>
<tr>
<td>Ductility — 1200°C (2192°F) (%)</td>
<td>5-8(a)</td>
<td>15-30(54)</td>
<td>15(54)</td>
</tr>
</tbody>
</table>

(a) Alpha/beta processed sheet.
conventional titanium alloys, although the room-temperature ductility and toughness levels of the titanium aluminides are always well below those of conventional titanium alloys.

Considering the aforementioned similarities between conventional alpha/beta titanium alloys and the alpha-two/beta-type titanium aluminides, it might be anticipated that parallels would also exist between the welding metallurgy of these alloy families. In the fusion welding of titanium alloys, several phenomena contribute to the room-temperature microstructure and resulting mechanical properties of the weldment. These phenomena include weld solidification behavior and its influence on the morphology of the prior-beta grains, beta decomposition in the weld fusion and heat-affected zones during continuous cooling and isothermal phase transformations during postweld heat treatment (Ref. 11). Of these phenomena, beta decomposition during continuous welding plays perhaps the greatest role in determining weld integrity and mechanical properties. Based on the above discussion, it would be anticipated that this phenomenon would similarly be critical in determining the weldability of alpha-two titanium aluminides. Preliminary fusion welding trials on Ti-14Al-21Nb have confirmed an influence of cooling rates on weld integrity and mechanical properties. As shown in Fig. 3, the brittleness and notch sensitivity associated with the microstructure near the crater of a rapidly cooled, autogenous gas tungsten arc (GTA) weld in Ti-14Al-21Nb are shown in Table 2. The thin sheet product was hot-rolled in the alpha plus beta phase field down to a thickness of 1.7 mm (0.067 in.) and air cooled to 650°C (1200°F), followed by a sub-beta-transus anneal at 1040°C (1900°F) for 1 h and fan cooled to room temperature. Figure 4A shows the coarse, prior-beta grain structure typical of beta-processed titanium and a transformed-beta microstructure comprised of coarse alpha-two platelets. Transmission electron microscope (TEM) bright-field analysis found the coarse alpha-two platelets to exhibit a relatively low dislocation density and also indicated the presence of thin beta strips at alpha-two platelet boundaries — Fig. 4B. Furthermore, selected-area diffraction (SAD) analysis and semiquantitative energy-dispersive x-ray analysis identified these strips as Nb-enriched beta phase.

The thin sheet product was hot-rolled in the alpha plus beta phase field down to a thickness of 1.7 mm (0.067 in.) and recrystallization annealed below the beta transus. The as-received microstructure exhibited equiaxed alpha-two grains with fine islands of beta phase at alpha-two grain boundaries — Fig. 5. Selected-area diffraction patterns shown in Fig. 5B confirmed the DO_{19} superlattice structure of the alpha-two and indicated an absence of ordering in the beta phase.

### Table 2 — Chemical Compositions of Ti-14Al-21Nb Base Metals (wt-%)

<table>
<thead>
<tr>
<th>Element</th>
<th>Ring Roll</th>
<th>Sheet</th>
</tr>
</thead>
<tbody>
<tr>
<td>Al</td>
<td>13.70</td>
<td>14.80</td>
</tr>
<tr>
<td>Nb</td>
<td>20.30</td>
<td>21.30</td>
</tr>
<tr>
<td>Fe</td>
<td>0.10</td>
<td>0.065</td>
</tr>
<tr>
<td>Si</td>
<td>0.085</td>
<td></td>
</tr>
<tr>
<td>O</td>
<td>0.065</td>
<td>0.058</td>
</tr>
<tr>
<td>N</td>
<td>0.007</td>
<td>0.006</td>
</tr>
<tr>
<td>H</td>
<td>0.002</td>
<td></td>
</tr>
<tr>
<td>Ti</td>
<td>bal.</td>
<td>bal.</td>
</tr>
</tbody>
</table>
Continuous-Cooling Phase Transformation Study

Continuous-cooling phase transformations (CCT's) in Ti-14Al-21Nb were investigated using dilatometry. Rectangular samples 3.2 x 6.3 x 76 mm (0.125 x 0.25 x 3.0 in.) were electro-discharge machined from the rolled ring, pickled in a solution of 5 mL HF + 40 mL HNO₃ + 55 mL H₂O and thermally cycled in a Gleeble 1500 system equipped with a °C-strain" dilatometer. The thermal cycle employed during dilatometry involved linearly heating to 1200°C (2192°F) (beta transus + 65°C (117°F)) in 3 s and holding at that temperature for 30 s to promote compositional homogenization. Average cooling rates from 150°C/s (270°F/s) to 10°C/s (18°F/s) between 1200°C (2192°F) and the transformation start temperature were achieved by free cooling, while slower cooling rates from 10°C/s (18°F/s) to 1°C/s (1.8°F/s) required controlled cooling. In addition, a specimen was thinned to 0.75 mm (0.03 in.) and water quenched at a rate of approximately 750°C/s (1350°F/s). Figure 6 shows temperature and dilation curves versus time for a specimen free cooled at a rate of 20°C/s (36°F/s). Although the phase transformation indications on the dilation curve corresponded well with the thermal arrest on the temperature curve, dilatometry was more effective in delineating the phase transformation completion temperature, particularly for specimens cooled at slower rates. All Gleeble testing was performed in an initial vacuum of 5 x 10⁻⁵ torr.

Generation of Simulated GTA Weld Fusion and Heat-Affected Zone Specimens

Simulated GTA weld fusion zone structures were produced using the Gleeble 1500 system by "superimposing" a transformed-beta microstructure on the coarse, columnar-shaped beta grain structure of an actual GTA weld. This procedure allowed the accurate and reproducible generation of weld structures representing a wide range of GTAW cooling rates. Autogenous, GTA welds, with depth of fusion through the sheet, were produced on 38 x 152 mm (1.5 x 6.0 in.) specimens (90 A, 10 V, 0.125-in. arc length, 22 L/s (25 cfh) argon shielding gas). GTA welds in all specimens were oriented longitudinal to the sheet rolling direction. The pickling of specimens just prior to welding and their welding in a Plexiglass chamber purged with high-purity argon assured low contamination and porosity-free welds. The welds were machined into specimens 12.7 X 76.2 mm (0.5 X 3 in.) and the weld reinforcement was removed by grinding. Subsequent Gleeble thermal simulation involved heating from room temperature to 1200°C (2192°F) in 5 s, holding at temperature for 30 s and cooling at rates of 50°C, 25°C, 10°C and 5°C/s (90°F, 45°F, 18°F and 9°F/s). The 50°C/s and 25°C/s cooling rates were achieved by free cooling, while the lower cooling rates were accomplished by a programmed cooling rate. It is important to recognize that the generation and evaluation of such "simulated" weld fusion zone specimens did not consider differences in fusion zone prior-beta grain size on the resulting transformed microstructure, mechanical properties and fracture behavior. Considering the extremely coarse prior-beta grain size exhibited by the GTA weldment fusion zone utilized in...
this simulation, in the context of previous studies of weld energy input on fusion zone grain size in alpha-beta titanium alloy GTA weldments (Ref. 11), it was not anticipated that the prior-beta grain sizes exhibited by actual GTA welds produced with the slower simulated weld cooling rates would increase significantly (Ref. 11).

Simulated weld HAZ specimens were generated from 12.7 × 76-mm (0.5 × 3-in.) coupons of the sheet in a manner similar to that used for the fusion zone specimens, except that no weld was present. However, for the simulated HAZ specimens, the hold time at 1200°C (2192°F) was extended to 90 s to allow beta grain growth to a size comparable to that observed in the near-HAZ of actual GTA welds. During the Gleeble generation of both simulated weld fusion and HAZ specimens, the specimen free span (i.e., jaw spacing) was selected such that a nearly constant peak temperature and cooling rate were maintained over a length of at least 12.7 mm (0.5 in.).

Metallurgical Characterization

Representative CCT and simulated FZ and HAZ specimens were sectioned from the hot zones of Gleeble specimens and mounted in Epomet. Metallographic preparation involved grinding from 240- through 600-grit silicon carbide (SiC), rough polishing on 6- and 3-micron diamond paste and final polishing with 0.06-micron colloidal silica. Microstructural details were revealed by etching with Kroll’s reagent comprised of 1-3 mL HF, 4-6 mL HNO₃ and 100 mL H₂O. Analysis of bulk metallographic specimens was principally light microscopy. Selected-diameter and simulated weld specimens were also examined using TEM. Samples were carefully removed by diamond sawing from the actual and simulated weld fusion zones and prepared by grinding on SiC paper to a thickness of 100 microns, followed by jet electropolishing to perforation in a solution of 250 mL methanol, 150 mL butyl cellulose, and 18-20 mL perchloric acid (60%) at −40°C. TEM examination and selected-area electron diffraction were performed on a JEOL 2000FX electron microscope operated at 200 kV. Semi-quantitative energy-dispersive x-ray analysis was performed on a Tracor-Northern 5500 system.

Mechanical testing of the simulated weld structures included microhardness testing, three-point guided-bend testing and tensile testing. Knoop microhardness testing was conducted on mounted metallographic specimens using a 50-g load. Three-point guided-bend tests were performed on simulated weld fusion and HAZ specimens, which were surface ground to a thickness of approximately 0.9 mm (0.035 in). Most specimens were ground perpendicular to the longitudinal speci- men axis using 60-grit abrasive. However, in order to examine the effect of notch sensitivity on bend ductility, an additional set of simulated weld fusion zone specimens was prepared with the surface ground parallel to the longitudinal specimen axis using a liner 120-grit abrasive. Two to four specimens of each type were deformed at progressively higher strain levels until crack initiation. Simulated weld fusion zone specimens were also machined into all-fusion-zone tensile specimens: 12.7 mm (0.5 in.) long × 3.2 mm (0.125 in.) wide × 1.7 mm (0.067 in.) thick and tested to fracture at an extension rate of 0.01 mm/s (0.0025 in./min).

Fracture surfaces of the bend and tensile tested specimens were examined using an ETEC Autoscan scanning electron microscope (SEM) operated at 20 kV.

Results

Continuous-Cooling Phase Transformations

Figure 7 shows the continuous-cooling phase transformation diagram generated by dilatometry for the Ti-14Al-21Nb rolled ring. As indicated, the BCC-to-HCP transformation start temperature for continuous cooling were appreciably lower than the equilibrium beta transus temperature, even at the slowest cooling rate of 1°C/s (1.8°F/s). Increases in cooling rate from 1°C/s to 150°C/s (270°F/s) were found to continuously decrease the transformation start temperature from approximately 950°C (1740°F) to 800°C (1470°F). Although the systematic generation of specimens exhibiting controlled cooling rates above 150°C/s was not possible with the Gleeble system, the water quenching of a dilatometry specimen at a rate of 750°C/s (1350°F) showed no dilution or thermal arrest, indicating the complete retention of beta phase down to room temperature.

Light micrographs of the continuously-cooled specimens are shown in Fig. 8. All specimens exhibited a coarse, nearly equiaxed prior-beta grain size with an average grain diameter of approximately 1-3 mm (0.04-0.12 in.). Beta grain boundaries were frequently decorated with semicontinuous alpha-two phase, which nucleated heterogeneously during weld cooling. As expected, the width and continuity of this grain boundary phase were greatest for the most slowly cooled specimens and continually decreased with increasing cooling rate.

Intragranular transformed-beta microstructures of specimens cooled at the slowest rates, i.e., 1°C/s (1.8°F/s) and 5°C/s (9°F/s), were comprised of coarse alpha-two platelets which varied greatly in size due to the “autopartitioning” nature of alpha-two platelet nucleation and propagation within the coarse beta grains. During cooling from the beta phase field, the initial platelets nucleated at beta grain boundaries (as indicated in Fig. 8A) and within the beta grains, and propagated unimpeded across the beta grains, thereby partitioning the beta grains. Subsequent platelets that formed were limited in length to the size of these partitioned regions and in forming further partitioned the remaining beta phase. As shown in Fig. 8B, this autopartitioning led to appreciable differences in alpha-two platelet lengths and widths within a single prior-beta grain. The irregular shapes commonly exhibited by the larger alpha-two platelets was also attributed to their impingement by smaller plates during this process. Another characteristic of the coarser alpha-two platelets was the presence of a distinct midrib located along their longitudinal centerline.

The bright-field TEM micrographs in Fig. 9 show the relatively coarse alpha-two platelets in a specimen cooled at 1°C/s (1.8°F/s). The alpha-two platelets (ordering was confirmed by SAD analysis) appeared somewhat rounded and occasionally exhibited a midrib. TEM bright-field imaging and SAD of the platelet boundaries further indicated the presence of retained-beta phase at these locations. EDS analysis found a significant enrichment of beta-stabilizing element Nb in the retained beta phase, suggesting beta decomposition at least in part by a diffusional mechanism.

Increased cooling rates promoted the
formation of morphologically similar, but continually finer and more acicular alpha-two platelets. At the highest cooling rates of 100°C/s (180°F/s) and 150°C/s (270°F/s), the acicular platelet microstructure was extremely fine and barely resolvable with light microscopy. Bright-field TEM micrographs in Fig. 10 of a specimen cooled at 100°C/s reveal an extremely fine platelet size. Although midribs could not be delineated within the fine platelets, the highly acicular nature of the transformation product and high internal dislocation density were consistent with the characteristics of alpha-prime martensite commonly observed in rapidly cooled alpha-beta titanium alloys (Ref. 13). SAD analysis of the fine platelet structure was unable to clearly resolve superlattice reflections associated with an ordered HCP structure. This absence may be explained by the presence of extremely fine domains or possibly transformation to a disordered martensitic product. TEM bright-field analysis and SAD further indicated an absence of retained-beta phase, which also suggested beta decomposition via a diffusionless/shear mechanism.

Consistent with the absence of an observed dilation or thermal arrest, the optical microstructure of the quenched specimen exhibited a coarse prior-beta grain structure with no evidence of transformed products.

Knoop microhardness values for the continuously cooled dilatometry specimens are shown in Fig. 11. It should be noted that hardness values within a single specimen varied by as much as 10–20 KHN due to crystallographic orientation effects, and that the values shown represent an average of five to ten measurements. As expected, the hardness increased with increasing cooling rate due to the formation of an increasingly finer transformed structure. The abrupt decrease in hardness for a cooling rate of 750°C/s (1350°F/s) was associated with retention of the soft, single-phase retained-beta microstructure.

Analysis of the GTA Weldment

Figure 12 shows a transverse section across the autogenous GTA weld in Ti-14Al-21Nb thin sheet and the corresponding hardness traverse. Macrostructural characteristics of the weld zone generally paralleled those observed in typical titanium alloy welds (Ref. 11). Coarse, columnar-shaped prior-beta grains were observed to nucleate epitaxially from coarse, equiaxed grains in the near-HAZ (i.e., the region that completely transformed to beta phase on heating). A close examination of these columnar prior-beta grains, particularly in heavily etched specimens, revealed evidence of a dendritic solidification structure. This structure would be expected to result from the segregation of Nb to dendrite cores (k > 1) and aluminum to dendrite interstices (k < 1) during solidification and an absence of complete compositional homogenization during cooling through the beta phase field. Electron microprobe analysis across this fusion zone by Cieslak (Ref. 14) has shown essentially no variation in Al and variations in Nb content of only 1–2 wt-% from the nominal level. In the adjacent near-HAZ, lower peak temperatures and shorter times above the beta transus temperature with increasing distance from the fusion line promoted a continually finer prior-
beta grain size. In contrast to the fusion zone and near-HAZ, the distinctive, dark-etching band associated with the far-HAZ structure was unique to welds in this titanium aluminide.

As indicated earlier, microstructural characteristics of titanium alloys are determined principally by cooling rates in the fusion and near-HAZ regions and heating and cooling rates in the far HAZ. The actual fusion zone thermal cycle shown in Fig. 13 was measured by plunging a platinum-platinum/rhodium thermocouple into the weld pool during welding, which indicated a cooling rate of approximately 65°C/s (117°F/s) between 1200°C (2192°F) and the transformation temperature of 870°C (1598°F). Consistent with this cooling rate, the fusion and near-HAZ regions exhibited fine, acicular alpha-two platelets which varied considerably in size, and also evidence of semicontinuous alpha-two at prior-beta grain boundaries — Fig. 14A. The TEM bright-field micrographs in Fig. 15 more clearly revealed the nature of the fusion zone microstructure, including the wide range of acicular platelet sizes and shapes, a blocky alpha-two structure along prior-beta grain boundaries and occasional midribs within the alpha-two platelets. Selected-area diffraction analysis by Cieslak (Ref. 14) has confirmed the ordering of these HCP platelets. It is important to note that light and transmission-electron microscopy revealed no apparent influence of fusion-zone microsegregation on the alpha-two platelet morphology.

Fig. 9 — TEM bright-field micrographs of Ti-14Al-21Nb dilatometry specimen cooled at 1 °C/s: A and B — alpha-two platelet structure, arrows indicate apparent midrib along alpha-two platelet centerline; C and D — parallel alpha-two platelets at increased magnification. Arrows in D indicate strip of beta phase along alpha-two platelet boundaries. A and B – 5000X; C – 10,000X; D – 20,000X.

Fig. 10 — TEM bright-field micrographs of Ti-14Al-21Nb dilatometry specimen cooled at 100 °C/s (180 °F/s). A – 10,000X; B – 40,000X.
Fig. 11 — Knoop microhardness versus cooling rate for dilatometry and simulated weld fusion and HAZ specimens in Ti-14Al-21Nb sheet.

Fig. 12 — Light micrographs (40X) and corresponding Knoop microhardness traverse for autogenous GTA weld in Ti-14Al-21Nb sheet. Cooling rate = 65 °C/s (117°F/s). Arrows indicate approximate fusion line.

Fig. 13 — Thermal cycle experienced in fusion zone of gas tungsten arc weld in Ti-14Al-21Nb sheet.

Figure 14B and C show microstructures of the near-HAZ and far-HAZ regions, respectively, at increased magnification. Although cooling rates in the near-HAZ would be expected to be slightly lower than those in the fusion zone, thereby producing a somewhat coarser alpha-two platelet structure, the HAZ alpha-two platelet structure actually appeared finer than that observed in the fusion zone. The dark-etching microstructure exhibited by the far-HAZ resulted from the unique manner in which the high-temperature beta phase consumed the equiaxed alpha-two grains during the heating portion of the weld thermal cycle. Growth of the beta phase appeared to occur both along the alpha-two grain boundaries and also linearly across the equiaxed alpha-two grains. Nearer to the fusion line, the width of these beta phase regions widened, ultimately transitioning to the near-HAZ. As indicated in Fig. 14C, details of the as-cooled microstructure present in this region could not be clearly delineated using light microscopy and are the subject of a continuing study using thin-foil electron microscopy techniques (Ref. 15).

As shown in Fig. 12, microhardness values across the GTA weld fusion and near-HAZ regions were appreciably greater than those observed in the far-HAZ and unaffected base metal. Also, the hardness in the near-HAZ was observed to be 40-50 KHN higher than that of the adjacent fusion zone, which was consistent with the finer alpha-two platelet structure observed in the former region. Consistent with the microhardness results, tensile testing found a fusion zone tensile strength of 1257 MPa (183 ksi), well above that of the base metal of 593 MPa (86 ksi). Despite special care in specimen preparation and tensile testing, inherent notch sensitivity of the microstructure promoted premature fracture at extensometer knife-edges or lightly scribed gauge length marks, or both, just after the onset of macroscopic plastic deformation, thereby precluding accurate measurements of tensile elongation and yield stress and likely causing reduced tensile strength. Progressive three-point bend testing of longitudinal-weld oriented specimens further confirmed the low overall ductility of the fusion zone and its notch sensitivity, with a total elastic plus plastic outer fiber bending strain of 1.1-1.2% at fracture for the specimen ground with 60-grit SiC perpendicular to the stress direction and 1.4-1.7% for specimen ground with 120-grit SiC parallel to the stress direction, respectively (Table 3).

SEM fractographic analysis of the GTA weld bend specimen showed a macroscopically faceted fracture surface in the coarse-grained fusion zone and near-HAZ regions, with the facet size and shape related to the morphology of the underlying prior-beta grains — Fig. 16. Fracture sur-
Fig. 14 – Light micrographs of regions in autogenous GTA weld in Ti-14Al-21Nb sheet: A – fusion zone; B – near HAZ; C – far HAZ. 1000X.

Fig. 15 – TEM bright-field micrographs of GTA weld fusion zone in Ti-14Al-21Nb sheet: A – acicular alpha-two platelets; B – arrow indicates apparent midrib along alpha-two platelet centerline; C – massive alpha-two at prior-beta grain boundary (indicated by arrows). A – 10,000X; B – 33,000X; C – 10,000X.
Macrostructural characterization of the simulated weld fusion zone specimens indicated that the original columnar prior-beta grain structure was essentially unaffected by the Cleeble thermal cycle. Heavy etching of specimens cooled at 5°C/s (9°F/s) and 10°C/s (18°F/s) exhibited coarse alpha-two platelets and semicontinuous alpha-two along prior-beta grain boundaries. Although bright-field TEM analysis of specimens cooled at 5°C/s showed a contrast at platelet boundaries indicative of beta phase, its presence could not be confirmed with SAD analysis. In addition, EDS analysis showed no evidence of Nb enrichment at these boundaries. Increased cooling rates promoted a refinement in the alpha-two platelet size and reduced the extent of alpha-two precipitation along prior-beta grain boundaries. A midrib structure along the centerline of alpha-two platelets was observed with light microscopy in each of the simulated weld microstructures. A direct comparison of the fusion zone and HAZ microstructures generated at identical cooling rates showed that the alpha-two platelet structure in the HAZ specimens generally appeared finer, with this difference being most apparent at the higher cooling rates.

Figure 11 provides a comparison of Knoop microhardness data obtained for the simulated weld zone specimens with that of specimens generated in the original CCT study. Consistently greater hardnesses of the simulated weld zone specimens versus the CCT specimens was attributed to differences in the chemical composition of these two product forms (Table 2). Hardnesses of the simulated weld HAZ specimens were generally greater than those observed in the simulated weld FZ specimens, particularly at the higher cooling rates. These observations were consistent with the finer alpha-two platelet structure observed in the HAZ region for comparable cooling rates.
and with hardness values obtained for the actual GTA weldments—Fig. 12.

Tensile strengths measured for the simulated fusion-zone specimens exceeded that of the base metal for all weld cooling rates. As shown in Table 3, tensile strength increased with increasing cooling rate from 944 MPa (137 ksi) for a specimen cooled at 5°C/s (9°F/s) to 1137 MPa (165 ksi) for a specimen cooled at 25°C/s (45°F/s). Although transverse-weld oriented tensile testing was not performed in the present study, hardness traverses across the actual GTA weldment indicated an absence of softening in the weld HAZ, which would suggest fracture of such specimens in the unaffected base metal.

As shown in Table 3, three-point bend testing of simulated weld fusion and HAZ specimens found the total plastic plus plastic outer-fiber tensile strain at fracture to be consistently low and to increase with a decrease in the weld cooling rate. Ductility levels at fracture for simulated fusion zone and HAZ specimens prepared by grinding with a 60-grit SiC abrasive parallel to the longitudinal specimen axis were nearly the same for identical cooling rates, ranging from a minimum of approximately 0.9% for HAZ specimens cooled at 50°C/s (90°F/s) to a maximum of about 1.5% for fusion zone and HAZ specimens cooled at 5°C/s (9°F/s). Grinding of the specimen surfaces in a direction perpendicular to the longitudinal specimen axis with a finer 120-grit SiC abrasive markedly increased the bend ductility values by 30 to 45%. A maximum bend ductility at failure of 2% for the simulated weld fusion zone specimen cooled at 5°C/s was just below that of a base metal bend specimen tested with the bending stress oriented transverse to the sheet rolling direction. It should be noted that the base metal sheet exhibited a ground surface comparable or slightly finer than that provided by grinding with 120-grit SiC.

SEM fractographic analysis of the fractured bend specimens showed a distinct influence of cooling rate on macroscopic and microscopic fracture characteristics. As shown in Fig. 16 for the actual GTA weld fusion zone (cooling rate = 65°C/s (117°F/s)) and Figs. 20A and B for a simulated weld HAZ specimen cooled at 50°C/s (90°F/s), fracture of rapidly-cooled specimens occurred transgranularly with the fracture surfaces exhibiting a macroscopically faceted appearance. In general, the facet size and shape correlated with the morphology of the underlying prior-beta grains. On a microscopic scale, fracture occurred by cleavage across the fine alpha-two platelets with limited ductile tearing, as shown in Fig. 16D. Decreasing the weld cooling rate promoted a macroscopically flatter, more fibrous-appearing transgranular fracture surface, as shown in Figs. 19A and 20C for simulated weld fusion zone and HAZ.

Table 3—Mechanical Properties of Simulated Weld Fusion and HAZ Specimens

<table>
<thead>
<tr>
<th>Specimen Type</th>
<th>Cooling Rate(a) °C/s (°F/s)</th>
<th>Knoop Hardness</th>
<th>Tensile Strength MPa (ksi)</th>
<th>Bend Ductility(b) %</th>
</tr>
</thead>
<tbody>
<tr>
<td>FZ(e)</td>
<td>5 (9)</td>
<td>337</td>
<td>944 (137)</td>
<td>1.3-1.5 (17-20)</td>
</tr>
<tr>
<td>FZ(e)</td>
<td>10 (18)</td>
<td>380</td>
<td>965 (140)</td>
<td>1.3-1.5 (14-17)</td>
</tr>
<tr>
<td>FZ(e)</td>
<td>25 (45)</td>
<td>390</td>
<td>1137 (165)</td>
<td>1.1-1.2 (14.7)</td>
</tr>
<tr>
<td>FZ(e)</td>
<td>50 (90)</td>
<td>415</td>
<td>—</td>
<td>1.1-1.2</td>
</tr>
<tr>
<td>FZ (AW)(e)</td>
<td>65 (117)</td>
<td>425</td>
<td>1261 (183)</td>
<td>1.1-1.2 (14.7)</td>
</tr>
<tr>
<td>HAZ(e)</td>
<td>3 (9)</td>
<td>334</td>
<td>—</td>
<td>1.3-1.6</td>
</tr>
<tr>
<td>HAZ(e)</td>
<td>10 (18)</td>
<td>386</td>
<td>—</td>
<td>1.1-1.3</td>
</tr>
<tr>
<td>HAZ(e)</td>
<td>25 (45)</td>
<td>425</td>
<td>—</td>
<td>0.9-1.1</td>
</tr>
<tr>
<td>HAZ(e)</td>
<td>50 (90)</td>
<td>450</td>
<td>—</td>
<td>0.9-1.1</td>
</tr>
<tr>
<td>BM (longitudinal rolling direction)</td>
<td>—</td>
<td>210</td>
<td>592 (86)</td>
<td>3.6-4.2</td>
</tr>
<tr>
<td>BM (transverse rolling direction)</td>
<td>—</td>
<td>210</td>
<td>—</td>
<td>2.0-2.5</td>
</tr>
</tbody>
</table>

(a) Cooling rate between 1200°C (2192°F) and transformation start temperature.
(b) Minimum outer fiber bending strain calculated from radius final die around which specimen was deformed without failure, maximum outer fiber bending strain represents subsequent die for which specimen failed. Fusion zone values in parentheses indicate bending strain values for specimens prepared by grinding with 120-grit SiC parallel to maximum tensile bending stress direction.
(c) Simulated weld fusion zone.
(d) As-welded weld fusion zone.
(e) Simulated weld heat-affected zone.

Fig. 17—Light micrographs of Ti-14Al-21Nb simulated weld fusion zone specimens cooled at: A—5°C/s (9°F/s); B—10°C/s (18°F/s); C—25°C/s (45°F/s); D—50°C/s (90°F/s). 1000X.
specimens, respectively, cooled at 5°C/s. On a microscopic scale, a decrease in cooling rate in both simulated fusion (Figs. 19B and C) and HAZ specimens (Figs. 20C and D) promoted a coarser scale of the fracture surface topography and an increased proportion of ductile tearing versus cleavage fracture.

Discussion

Continuous-Cooling Phase Transformations

The Ti₃Al-Nb pseudo-binary phase diagram shown in Fig. 2 is useful for describing phase transformations in the Ti-14Al-21Nb titanium aluminide under equilibrium conditions. On cooling, the high-temperature BCC beta phase transforms to the low-temperature HCP alpha-two phase at the "beta transus" temperature, which is approximately 1135°C (2075°F) for Ti-14Al-21Nb. As cooling proceeds, proportions of the alpha-two and beta phases are dependent entirely on the phase equilibrium at a particular temperature.

The results of Gleeble dilatometry and corresponding microstructural characterization have shown that beta decomposition in Ti-14Al-21Nb is appreciably more complex under conditions of continuous cooling. As shown in Fig. 7, the cooling rate significantly influences the temperature at which beta decomposition begins. Even at a relatively slow cooling rate of 1°C/s, the transformation start temperature was observed to be approximately 175°C (315°F) below the equilibrium beta-transus. This appreciable suppression of the transformation start temperature may be the result of the low diffusivity of Nb in the beta phase in precluding effective nucleation and growth of the HCP phase. Increased cooling rates continuously decreased the transformation temperature, with total suppression of beta decomposition observed in a specimen quenched at 750°C/s (1350°F/s). The complete suppression of beta decomposition during the rapid quenching of Ti-14Al-21Nb has also been reported in the fusion zone of Nd:YAG laser beam welds (Ref. 16). TEM/SAD analysis of the laser beam weld structure indicated an ordering of the beta phase during cooling to a B2 (CsCl) superlattice. It should be noted that the transition from 100% martensitic alpha-two to B2 beta phase with increasing cooling rates is gradual. Characterization of Ti₃Al + Nb alloy ribbons by Kaufman (Ref. 17) showed mixtures of these two phases, with greater proportions of alpha-two in thicker, more slowly cooled regions of the ribbon.

It is of interest to note that cooling rate studies by Jepson, et al. (Ref. 18), on Ti-5 to 17.5 at.-% Nb alloys showed a similar decrease in the beta-to-alpha transformation temperature with increased cooling rate with a total suppression of this transformation at sufficiently high rates. In contrast to the present study, however, his results showed a discontinuity in the cooling curve (a drop in transformation temperature with increasing cooling rate), which was interpreted as a boundary between the diffusional nucleation and growth of alpha at slow cooling rates and transformation by an isothermal martensitic transformation at higher cooling rates. The absence of an observable discontinuity in the Ti-14Al-21Nb CCT curve obtained in this study suggests that it may occur at a cooling rate below 1°C/s or that a distinct change from a martensitic to a diffusional transformation does not occur in this alloy.

A comparison of the continuously cooled specimens considered in the present investigation showed morphologically similar microstructures comprised of randomly oriented alpha-two platelets, which became progressively finer and more acicular with increasing cooling rate. Considering this gradual change in platelet morphology and the continuous nature of the CCT curve over this range of cooling rates, it is anticipated that these microstructures were generated by the same basic phase transformation mechanism. The presence of a midrib structure within the alpha-two platelets, which has been observed within coarse martensite plates in continuously cooled Ti-6Al-2Nb-1Ta-0.8Mo (Ref. 19), strongly suggests transformation by a diffusionless/shear transformation mode. TEM examination of CCT specimens cooled at rates down to 5°C/s (9°F/s) and of the GTA weld fusion zone, which showed an apparent absence of retained beta phase, further indicates a diffusionless/shear transformation mode at these cooling rates. Based on these observations, it may be suggested that the continuous cooling curve shown in Fig. 7 actually describes a range of M₇₆ temperatures.

The observation of thin, Nb-enriched beta phase ribbons between alpha-two platelets in the specimen cooled at 1°C/s may indicate a transition from a completely diffusionless/shear transformation to one involving diffusion. The absence of an observable discontinuity on the cooling curve suggests that this change in phase transformation behavior is not abrupt, and that a gradual transition occurs in which beta decomposition involves both diffusionless/shear and diffusional growth mechanisms. Considering the platelet mor-
Phology of the alpha-two and the presence of a midrib structure, it is suggested that beta decomposition initially involved a diffusionless/shear transformation. Once nucleated, platelet coarsening occurred by a diffusional growth mechanism that promoted the partitioning of Nb to the remaining beta phase and its stabilization. A similar martensitic growth followed by diffusional phase transformation has been observed in beta decomposition of binary Ti-Mo alloys (Ref. 20). It is important to note that this observed alpha-two platelet plus beta microstructure may also be explained by an initial shear transformation completely to an alpha-two platelet structure with the subsequent formation of beta phase at platelet boundaries via an "auto-tempering" reaction during slow cooling. Considering the continuous nature of the beta phase versus the presence of discrete particles, the former transformation mechanism appears more likely.

Limited SAD analysis performed in this study indicated that cooling rate also influences the ordering reaction in the HCP platelets. Studies by Sastry and Lipsitt (Ref. 4) on Ti3Al/Nb alloys suggested that ordering observed in specimens cooled at rates up to 65°C/s (117°F/s) occurred in the martensitic, diffusional HCP structure, or both, rather than transformation of beta directly to the ordered alpha-two. For specimens cooled at high rates (e.g., 100°C/s (180°F/s)), which transform at low temperatures, ordering may occur on an extremely fine scale or possibly be completely suppressed. A similar behavior has been observed by Sastry and Lipsitt (Ref. 4) in a Ti3Al-Nb alloy quenched from the beta phase field.

Although the present study has provided a basic understanding of how cooling rate influences weld zone microstructures in Ti-14Al-21Nb, it is apparent that the precise mechanics of beta decomposition and ordering reactions are complex and will require considerable detailed study before they can be completely understood.

Structure/Property/Fracture Relationships

Relationships between microstructures, mechanical properties and fracture characteristics of the continuously cooled Ti-14Al-21Nb microstructures generally paralleled those of conventional alpha-beta titanium alloy weldments, with the principal difference being the appreciably lower ductility exhibited by the titanium aluminide. The generation of increasingly finer, randomly oriented alpha-two platelets with increasing cooling rate resulted in an increased hardness and tensile strength. On a microscopic scale, these increases resulted from a decrease in alpha-two platelet size and possibly the anti-phase domain size within these platelets and an
increase in the dislocation density within the platelets. As is generally observed in conventional near-alpha and alpha-beta titanium alloys, the transformed-beta fusion zone and HAZ microstructures in Ti-14Al-21Nb exhibited hardeness in Ti-14Al-21Nb exhibited hardennesses and strengths well above those of the annealed base metal (i.e., joint efficiencies equaled 100%), even for the lowest cooling rate. The finer transformed-beta microstructures and greater hardennesses observed for the simulated near-HAZ specimens versus the simulated fusion zone specimens cooled at identical rates most likely resulted from subtle effects of minor, local compositional variations within the fusion zone in influencing alpha-two platelet nucleation and growth. Such segregation effects may, in part, explain why the differences were most noticeable for higher cooling rate welds in which a greater extent of segregation would be retained. An additional explanation for these structure and hardness differences may be related to differences in the prior-beta grain size of these weld regions.

Based on the above observations, it is apparent that ductility represents a mechanical property of critical concern in the welding of Ti-14Al-21Nb. From a morphological standpoint, the presence of relatively fine, randomly-oriented alpha-two platelet structure is beneficial in minimizing and dispersing planar slip within the alpha-two phase. In addition, fine beta strips between alpha-two platelets have also been shown to disperse slip and promote crack bifurcation and blunting. In the present study, the simulated weld fusion zone structure cooled at 5°C/s (9°F/s) exhibited the greatest ductility. Although TEM analysis could not identify beta strips between alpha-two platelets in this structure, the higher ductility and the observation of increased ductile tearing on the fracture surface of this specimen suggest the presence of beta phase. In comparison, the lower ductilities and higher proportions of faceted cleavage fracture in specimens cooled at higher cooling rates resulted from an absence of beta phase in these microstructures.

It is of interest to note that despite the presence of alpha-two along prior-beta grain boundaries, particularly in the slowly cooled structures, fracture consistently occurred transgranularly. This absence of intergranular fracture likely accounted for negligible differences in bend ductility between the identically cooled fusion zone and HAZ specimens, despite significant differences in prior-beta grain size and morphology.

Additional Weldability Considerations

Limited Varestraint testing of cast Ti-14Al-21Nb found solidification and liquation cracks nearly impossible to produce, even at high augmented strain levels (Ref. 12). As illustrated in Fig. 3, however, the low room-temperature ductility and notch sensitivity of these alloys can make them susceptible to cold cracking during welding. Such solid-state cracking may be minimized by maintaining a preheat and minimum interpass temperature during welding 1) reduce the weld cooling rate and thereby provide improved weld zone ductility and 2) stress relieve the specimen and thereby minimize cold cracks initiated at weld discontinuities. In addition, the decrease of weld discontinuities, which may serve as stress risers, will also reduce tendencies for crack initiation.

Postweld heat treatment is an additional important consideration regarding the weldability of alpha-two titanium aluminides. Since these alloys are designed principally for application at elevated temperatures, postweld heat treatment will generally be required both for stress relief and for microstructure stabilization. A preliminary study (Ref. 12) has shown microstructure stability at temperatures above about 600°C (1112°F) to be highly dependent on the cooling rate experienced during welding. Coarser alpha-two platelet structures associated with slow cooling rates are stable to temperatures exceeding 700°C (1292°F), while the extremely fine structures associated with rapid cooling rates can be very unstable at temperatures below 500°C (932°F). In the fine alpha-two microstructures, the instability is manifested by the autorecrystallization of fine alpha-two platelets along prior-beta grain boundaries, which promotes the formation of a coarse, undesirable alpha-two morphology at these boundaries. Based on these initial heat treatment results, the use of relatively slow welding rates would be recommended to optimize microstructure stability. Such microstructure stability would be of critical importance in the joining of large structures for which postweld heat treatment is not feasible.

Implications of CCT Study on Joining Ti-14Al-21Nb Using Alternate Welding Processes

Results of the CCT studies performed here can be utilized to predict the response of Ti-14Al-21Nb to joining via alternate welding processes. Figure 21 illustrates the macroscopic cross-section of a complete joint penetration electron beam weld produced in 6.3-mm (0.25-in.) thick Ti-14Al-21Nb plate machined from the rolled ring (120 kV, 7 mA, 12.6 mm/s (30 in./min)). Considering the high cooling rates experienced during electron beam welding, on the order of 10^3 to 10^4°C/s, the CCT curve in Fig. 7 would predict retention of the beta phase structure and formation of a relatively soft fusion zone. Indeed, examination of the weld fusion zone revealed a columnar beta grain macrostructure with no evidence of a transformed microstructure. In
addition, the hardness traverse shows a relatively low hardness across the fusion zone quite comparable to that of the water-quenched CCT specimen (KHN 330).

Inertia friction welding is a solid-state process widely used by the aerospace industry in the joining of similar and dissimilar alloys. Cooling rates experienced during the friction welding of titanium would be expected to range from about $10^2$ to $10^3 \degree C/s$, depending on the weld size and welding parameters. Figure 22 shows the transverse section of an inertia friction weld produced between 22.2-mm (0.875-in.) diameter rods machined from the rolled ring using the following welding conditions: moment of inertia —0.167 kgm$^2$ (3.93 lb-ft$^2$), rotational speed — 628 radians/s (6000 rpm), axial thrust load — 38.7 kN (8731 psi). Examination of the macroscopically featureless weld interface region at increased magnification revealed an extremely fine prior-beta grain structure, which was produced by dynamic recrystallization during high temperature deformation, and an extremely fine transformed-beta microstructure. These microstructural characteristics were consistent with the appreciable increase in hardness across the weld zone exceeding KHN 500. A correlation of the CCT specimen hardness data with inertia friction weld hardness suggests an actual weld cooling rate in the vicinity of 150$^\circ$ to 200$^\circ$C/s (270$^\circ$ to 360$^\circ$F/s). Based on the previously discussed structure/property/fracture relationships, failure of the inertia friction weld would be expected at low ductility by a brittle, cleavage mode of fracture. Indeed, fracture of a transverse-weld oriented specimen during three-point bend testing occurred along the weld interface with negligible ductility. SEM examination of the fracture surface revealed the macroscopically faceted, cleavage-type fracture observed previously as described for rapidly cooled microstructures — Fig. 23.

Conclusions

1. Continuous-cooling phase transformation studies in Ti-14Al-21Nb using dilatometry showed a continuous decrease in the BCC-to-HCP transformation start temperature with increasing cooling rate from 1$^\circ$ to 150$^\circ$C/s (1.8$^\circ$ to 270$^\circ$F/s). Water quenching at a cooling rate of 750$^\circ$C/s (1350°F/s) caused complete retention of the beta phase down to room temperature.

2. Continuous cooling of the Ti-14Al-21Nb from above the beta transus temperature at rates from 1$^\circ$ to 150$^\circ$C/s (1.8$^\circ$ to 270$^\circ$F/s) promoted the formation of acicular alpha-two microstructures with the platelet size decreasing with increasing cooling rate. Specimens cooled at a rate of 1$^\circ$C/s (1.8°F/s) exhibited both a midrib structure within the alpha-two platelets and Nb-enriched retained beta phase at platelet boundaries, thereby suggesting beta decomposition by a mixed shear/diffusional mode. An increasingly acicular alpha-two platelet morphology and absence of observable retained beta phase in specimens cooled at rates from 5$^\circ$C/s (9°F/s) to 150$^\circ$C/s (270$^\circ$F/s) indicated beta decomposition by a diffusionless/shear mechanism.

3. Simulated gas tungsten arc weld fusion and near-HAZ specimens exhibited microstructures comparable to those of dilatometry specimens. Increases in cooling rate were observed to increase hardness and tensile strength and decrease the ductility of these weld regions. Bend ductility levels were consistently low (below 2%) for all cooling rates. An observed influence of surface condition on fusion zone bend ductility indicated a notch sensitivity of the microstructures. Simulated weld regions produced at a cooling rate of 5$^\circ$C/s (9°F/s) exhibited the optimum combination of mechanical properties.

4. Fracture surfaces of the simulated weld fusion and HAZ microstructures produced at high cooling rates exhibited a faceted appearance, which was microscopically comprised of cleavage and duc-

![Fracture surface of three-point bend specimen of inertia friction weld in Ti-14Al-21Nb. A — 50X; B — 500X.](image-url)
tile tearing. Slower cooling rates promoted a macroscopically flat fracture, which exhibited an increasing proportion of ductile tearing.

5. Basic continuous-cooling transformation diagrams can be utilized to predict microstructure characteristics of welds produced by alternate joining processes, notably electron beam and inertia friction welding.

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References


