The Influence of Microstructures on the Notched Tensile Fracture of Ti–6Al–6V–2Sn Welds at Elevated Temperatures

Leu-Wen TSAY,1 Chung-Liang HSU1) and Chun Chen2)

1) Institute of Materials Engineering, National Taiwan Ocean University, Keelung 202, Taiwan, R.O.C.
E-mail: b0186@mail.ntou.edu.tw (L.-W. Tsay) 2) Department of Materials Science and Engineering, National Taiwan University, Taipei 106, Taiwan, R.O.C.

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KEY WORDS: notched tensile test; Ti–6Al–6V–2Sn; post-weld heat treatment.

1. Introduction

Titanium alloys of α+β type can be subjected to thermal or thermo-mechanical processes to achieve different combinations of strength and toughness for various industrial applications.1) The mechanical properties and fracture behavior of α+β titanium alloys are strongly related to the microstructures of the material.2) In general, a coarse acicular structure obtained by beta heat treatment has higher fracture toughness but considerably lower ductility than the equiaxed structure.2–4) It is reported that heat treatment of Ti–6Al–4V in the α–β field produces a marked increase in toughness with a slight loss in ductility relative to the mill-annealed condition.4) A lack of deformation compatibility at the interfaces between the α and β phases caused interfacial separations in the MB specimen as well as void formation in the W-704 specimen, leading to the low NTS of these specimens at elevated temperatures.

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2. Material and Experimental Procedures

Ti–6Al–6V–2Sn in sheet form with a thickness of 4.0 mm was used in the experiment. The chemical composi-
tion of the alloy in weight percent was 5.50Al, 5.40V, 1.90Sn, 0.5Fe, 0.54Cu, 0.02C, 0.011N, and a balance of Ti. The mill-annealed base metal (MB) specimen was heat treated at 732°C for 1 h, cooled in the furnace to 482°C and air cooled to room temperature. The microstructures of the MB specimen consisted of a low percentage of β at the boundaries of elongated α grains. Tensile properties of the MB specimen include an ultimate tensile strength of 1 117 MPa, a yield strength of 1 086 MPa and an elongation of 20%.

A CO2 laser was utilized for bead-on-plate welding in the keyhole mode in one pass at room temperature. The welding parameters included a laser power of 2.7 kW and a travel speed of 800 mm/min. A copper mirror with focal length of 200 mm was utilized for laser welding of Ti-6Al–6V–2Sn specimens. The focal point of laser beam was positioned at 0.5 mm below the specimen surface. All specimens were welded in the mill-annealed condition with the welding direction normal to the rolling direction. The welded specimen in the as-welded condition was named as the AW specimen. PWHTs were performed on the AW specimen at 482 and 704°C for 3 h in vacuum and then followed by argon-assisted cooling to room temperature; such specimens were designated as the W-482 and W-704 specimens, respectively.

Double-edge notched specimens,17,18 which had a thickness of 3.5 mm, width of 20 mm and notch depth of 7.0 mm (notch angle of 60° and notch radius of 100 μm), were used to determine the NTS values of the welds. The use of such configuration could completely reflect notch brittleness of the tested specimens under tensile loading.16–18 For the welds, the notches were located at the center of the fusion zone (FZ) to ensure the crack growth along the weld centerline during tensile loading. Notched tensile tests were carried out in air at a constant displacement rate of 1.0 mm/min. For high-temperature tests, the specimens were placed into a furnace and heated to the predetermined temperature (150, 300 and 450°C), then held at that temperature for 30 min before testing. The NTS results are presented as the average of at least three specimens for each testing condition. The fracture surfaces of distinct specimens were examined using a scanning electron microscope (SEM), with attention paid to the change in fracture features. The detailed microstructures of the specimens were examined with a transmission electron microscope (TEM). In addition, the fractured specimens were prepared for further examinations using an SEM to correlate the crack growth path with the microstructures.

3. Results and Discussion

The low energy input of the laser welding process resulted in a narrow fusion zone (FZ) as well as a heat-affected zone (HAZ) in the Ti–6Al–6V–2Sn weld. Owing to the uneven heat distribution during welding, the fusion zone (weld metal) exhibited the shape of a wine-glass cup with the top and bottom weld widths of about 3.8 and 2.7 mm, respectively. In the as-welded (AW) condition, the micro-hardness of the FZ could be as high as Hv 440, which was much harder than the mill-annealed base metal (MB) hardness of Hv 340. After a PWHT at 482°C, the peak hardness in the FZ of the W-482 specimen could reach Hv 460. A slight decline in FZ hardness to Hv 430 was found for the W-593 specimen. In the case of the W-704 specimen, a significant decrease in the FZ hardness was noticed; however, its hardness (Hv 380) was still higher than the MB (Hv 340). It was noted that the FZ was considerably harder than other regions in the weld, regardless of the PWHT conditions.

SEM micrographs showing the microstructures of various specimens are given in Fig. 1. Elongated α and intergranular β phases were seen in the MB specimen (Fig. 1(a)). The FZ in the AW condition consisted of coarse columnar structures with fine needle features in the matrix (Fig. 1(b)). After PWHT at 482°C, thin α films were found to decorate the grain boundaries of the W-482 specimen (Fig. 1(c)). For PWHTs at 593°C and above, coarsening of acicular α and thickening of α films at the grain boundaries was observed (Fig. 1(d)). Obviously, the grain boundary α became thicker with increasing PWHT temperature. TEM observations of the as-welded FZ revealed very fine acicular α uniformly distributed in a β matrix in the form of basket-weave structures (Fig. 2(a)). The high hardness of the as-welded FZ, as compared to the banded structure of the MB specimen, could be attributed to the presence of fine acicular α structures. Moreover, some of the grain boundaries contained island-like α precipitates (Fig. 2(a)). For the W-482 specimen, the intragranular and intergranular α (Fig. 2(b)) were coarser than those in the as-welded FZ (Fig. 2(a)). The slight increase in hardness in the W-482 specimen in comparison with the AW specimen was related to the decomposition of few retained β to α + β during the PWHT.19 The coarsening of α + β structures in the matrix and the formation of thick α layer at grain boundaries could account for the rapid decline in hardness of the W-704 specimen (Fig. 2(c)).

Figure 3 shows the results of the notched tensile tests at room temperature and 450°C. The NTS of the MB specimen (1 220 MPa) was a little higher than its ultimate tensile strength (1 117 MPa) at room temperature. As shown in Fig. 3(a), the NTS values of the welds were noticeably lower than that of the MB specimen, and the W-482 specimen also showed high notch brittleness at room temperature. Besides, the W-704 specimen had a higher NTS.
(1060 MPa) than the AW (977 MPa) or W-482 (818 MPa) specimens at room temperature. This indicated that an improper PWHT caused notch brittleness of the weld at room temperature. In contrast to room temperature behavior, the MB specimen exhibited a remarkable drop in NTS and a great improvement in ductility at 450°C, as shown in Fig. 3(b). The W-704 specimen also showed a slight decline in NTS and an improved ductility at 450°C. The AW and W-482 specimens tested at 450°C seemed to resist softening with a slight increase in ductility. At 450°C, the W-482 specimen exhibited the highest NTS among the specimens being evaluated.

The results of notched tensile tests of distinct specimens at different temperatures are shown in Fig. 4. The MB specimens exhibited a rapid decrease in NTS with increasing temperature; it had the highest NTS at room temperature but the lowest NTS at 450°C among the specimens. All welds in the tests had a trend of increasing NTS from room temperature to 150°C, but a gradual decrease in NTS with further rises in temperature. At test temperatures of 300 and 450°C, the AW and W-482 specimens had similar NTS values, which were much higher than those of the MB and W-704 specimens. The results revealed that the AW and W-482 specimens with high hardness at room temperature were susceptible to notch brittleness, however, these specimens showed high NTS and improved ductility at elevated temperatures. Such characteristics could be related to the microstructures and fracture behavior of the specimens, and were discussed later in the text.

Figure 5 displays the micro-hardness of the specimens after notched tensile tests, in which the measurements were taken in the region very close to the fracture locations. The hardness of the MB and W-704 specimens increased with increasing the test temperature. This confirmed that plastic deformation resulted in strain-hardening of the specimens before rupture at elevated temperatures. For the W-482 specimen, a limited variation in hardness after testing at different temperatures could be attributed to the inherent stable microstructures with high hardness over the testing temperature range. The W-482 specimen was sensitive to notch brittleness at room temperature, but improved ductility at elevated temperatures facilitated notch-blunting and reduced the notch brittleness of the W-482 specimen. In the case of the AW specimen, a gradual rise in hardness could be as a result of strain hardening and the response to age-hardening of the alloy.

Figure 6 shows the macro-fracture appearance of the specimens tested at room temperature and 450°C. With in-
creasing test temperature, an increased range of the slant fracture for the MB specimen was observed (Fig. 6(a)). This is indicative of the ease of deformation of the material at elevated temperatures. At room temperature, the AW and W-482 specimens exhibited extensive flat fracture, indicating the essentially brittle nature of these specimens (Figs. 6(b) and 6(c)), especially for the W-482 specimen. For the W-704 specimen (Fig. 6(d)), the wide slant fracture corresponded to lowered notch brittleness or improved NTS of the specimen at room temperature. The improved NTS of the AW and W-482 specimens (Fig. 4) at elevated temperatures was confirmed by a notable increase in the shear fracture regions (Figs. 6(e) and 6(f)).

Figure 7 presents SEM fractographs showing the typical fracture appearance of various specimens. The MB specimen had extensive transgranular dimple fractures both at room temperature and at 450°C (Fig. 7(a)), with a larger dimple size at 450°C. The fracture modes of the AW specimen were affected by the crack growth direction relative to the solidification structures. A predominantly transgranular fracture with fine dimples was observed for the crack growth normal to the solidified structure. However, few interdendritic separations with solidified features were found for the crack growth along the solidification direction at room temperature (Fig. 7(b)). Such interdendritic separations were not observed in the AW specimen tested at/above 300°C and not in the specimens with PWHT. It was deduced that interdendritic separations could be related to the presence of some retained β at the columnar boundaries of the AW specimen. The decomposition of residual β at the grain boundaries altered the fracture characteristics at elevated temperatures. Extremely fine and shallow dimples were observed in the W-482 specimen, which could be attributed to the uniform distribution of fine acicular structure. Moreover, dimples mixed with grain boundary shear were observed in the W-704 specimen tested at room temperature (Fig. 7(c)). The presence of thick grain boundary α layer was believed to promote grain boundary shear in the W-704 specimen during the notched tensile test. With increasing test temperature, coarse dimples with an increased extent of grain boundary shear occurred in the AW and W-482 specimens. This indicated that grain boundary shear was activated at elevated temperatures for the AW and W-482 specimens. Such observations also confirmed that the grain boundary was weak relative to the grain interior of the specimens. Moreover, a significant increase in intergranular fracture along columnar boundaries occurred for the W-704 specimen tested at 450°C (Fig. 7(d)).

Figure 8 shows the correlation of the crack growth path and the associated microstructure near the fracture region. This indicated that the β phase became elongated after straining, and the crack tended to grow along the α/β interfaces in the MB specimen, particularly at high temperatures (Fig. 8(a)). This also demonstrated the fact that the lack of
deformation compatibility between the $\alpha$ and $\beta$ phases resulted in cracking at the interfaces in the MB specimen during straining. It was clear that the coarse $\beta$ in the MB specimen did not show a strengthening effect, especially at elevated temperatures. This can be seen in Fig. 4, where the MB specimen exhibits an obvious decline in NTS with increasing temperature. An increased $\beta$ content in the mill-annealed $\alpha+\beta$ alloy, e.g. the Ti–4.5Al–3V–2Fe–2Mo, results in a further drop of NTS at elevated temperatures.\textsuperscript{21} This could be related to the increased cracking routes of the $\alpha/\beta$ interfaces. In contrast, the fine acicular structures in the form of basket-weave in all welds were less likely to form interfacial separations and exhibited higher NTS values than the MB specimen at elevated temperatures.

The fracture morphology at the grain boundaries should depend on the discrepancy in strength/hardness between the boundary and grain interior regions. The presence of thick grain boundary $\alpha$ layer in the W-704 specimen was anticipated to facilitate grain boundary shear/deformation at room temperature. The fracture appearance of the W-704 specimen was characterized by transgranular dimples with shear features at grain boundaries (Fig. 7(c)). Therefore, the grain boundary $\alpha$ was helpful to lower the notch brittleness of the specimen at room temperature. For the W-704 specimen tested at 450°C, a curved grain boundary (Fig. 8(b)) demonstrated that the weak grain boundary was easier to deform than the strong matrix. The mismatch strain concentrated at the grain boundaries\textsuperscript{22} is more likely to induce cracking thereafter. Therefore, the obvious drop in NTS of the W-704 specimen at 450°C was attributed to the presence of the grain boundary $\alpha$ layer. Besides, the formation of fine voids at the columnar boundaries accounted for the intergranular fine dimple fracture of the W-704 specimen at 450°C (Fig. 7(d)). The lowered NTS of the W-704 specimen at 450°C was also attributed to the formation of fine voids at the columnar boundaries. Such grain boundary voids were hardly to be observed in the W-704 specimen tested at room temperature. It was concluded that an enhanced grain boundary shear, which was associated with the grain boundary $\alpha$ layer, alleviated the notch brittleness of the W-704 specimen at room temperature but lowered its NTS at 450°C.

4. Summary

With the presence of sharp notches, the maintenance of high notched strength at elevated temperature is required for the material to resist cracking and/or embrittlement under tensile straining. The notch sensitivity of Ti–6Al–6V–2Sn laser welds was determined by using double-edge notched specimens to confine fracture within the narrow fusion zone. The results revealed the NTS values of the AW and W-482 specimens were much lower than that of the MB specimen at room temperature; however, the trend was altered at elevated temperatures. The W-704 and MB specimens exhibited a noticeable decline in NTS at elevated temperatures, especially for the MB specimen at 450°C. In contrast, the improved ductility at 450°C could account for the increased NTS or reduced notch brittleness of the AW and W-482 specimens with increasing test temperature. The presence of thick grain boundary $\alpha$ in the W-704 specimen promoted grain boundary shear/deformation and therefore reduced the notch brittleness at room temperature. The typical fracture appearance of the W-704 specimen at room temperature was transgranular dimples accompanied by grain boundary shear. The observation of curved grain boundaries in the W-704 specimen implies that the grain boundary was weak relative to the grain interior at elevated temperatures. Consequently, the accumulated strain at the grain boundaries was more likely to induce cracking or void formation at elevated temperatures. The lack of deformation compatibility between the $\alpha$ and $\beta$ phases enhanced interface separations in the MB specimen or induced voids in the W-704 specimen, resulting in lowered NTS at elevated temperatures.

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