**Notched Tensile Strength and Fatigue Crack Growth Behavior of Ti-6Al-6V-2Sn Laser Welds**

Leu-Wen Tsay¹,*, Chun-Xian Lee¹ and Chun Chen²

¹Institute of Materials Engineering, National Taiwan Ocean University, Keelung 202, Taiwan, R.O. China
²Department of Materials Science and Engineering, National Taiwan University, Taipei 106, Taiwan, R.O. China

Notched tensile and fatigue crack growth tests were performed on Ti-6Al-6V-2Sn laser welds which were subjected to post-weld heat treatments (PWHTs) at various temperatures. In comparison with the mill-annealed base metal (BM), Ti-6Al-6V-2Sn laser welds showed notch brittleness at different degrees. Generally, the weld metal (WM) had a lower fatigue crack growth rates (FCGRs) than the BM. Fine acicular α distributed uniformly in the β matrix could account for the high hardness of the as-welded WM relative to the BM. With the PWHTs at higher temperatures, the coarsened microstructures of the welds reduced not only the hardness but also the sensitivity to notch brittleness. The fatigue fracture morphology of the WM was much rougher and irregular than that of the BM. A tortuous crack path in the WM as compared to the relatively straight path in the BM resulted in reduced FCGRs in the WM, particularly for the welds subjected to PWHTs at 704 °C or higher.

(Received January 6, 2009; Accepted April 23, 2009; Published June 25, 2009)

**Keywords:** notched tensile strength, fatigue crack growth rate, Ti-6Al-6V-2Sn, laser welding, post-weld heat treatment

1. Introduction

With the high strength-to-weight ratio combined with excellent corrosion resistance, titanium alloys have been widely applied to aerospace, marine, and chemical industries. In α + β titanium alloys, a wide range of mechanical and physical properties can be achieved by the control of microstructures, which depend both on processing history and heat treatments. Ti-6Al-6V-2Sn, an α + β titanium alloy, is designed to achieve an ultimate tensile strength of 1200 MPa and improve heat treatability in comparison with Ti-6Al-4V.1) At the same strength level, the Ti-6Al-6V-2Sn alloy has the advantages of higher ductility and fracture toughness over the Ti-6Al-4V alloy.¹

For industrial applications, joining the metals by welding is efficient and cost-effective means of fabricating complex structures.²–⁴) Nowadays, precision welding of titanium alloys with good quality can be made by laser welding.⁵,⁶) In the as-welded and heat-treated conditions, Ti-6Al-4V welds exhibit superior fracture toughness,⁷,⁸) and lower fatigue crack growth rates (FCGRs) than the mill-annealed Ti-6Al-4V specimens.⁸–¹⁰) It is known that the microstructure affects significantly the performance of titanium alloys. For example, the microstructures in plate forms possess a better resistance to crack propagation than microstructures (α + β processed) containing equiaxed α grains.¹⁰) It is reported that the coarse structure in the weld metal (WM) of Ti-6Al-6V-2Sn welds often results in lowered ductility and toughness.¹¹,¹²) In the Ti-6.5Al-1.9Zr-0.25Si alloy, the poor fracture toughness of electron beam welds is attributed to the poor energy absorbing capacity of thin α′ and α in the WM.¹³) The fracture toughness and the stress corrosion resistance of Ti-6Al-6V-2Sn welds can be improved by decreasing cooling rates or α′ content.¹⁴) With an appropriate post-weld heat treatment (PWHT), Ti-6Al-6V-2Sn welds may possess good ductility and fracture toughness.¹⁴) The higher fracture toughness and ductility of Ti-6Al-6V-2Sn welds can be obtained with the microstructures having coarser α plates in the matrix and at the grain boundaries.¹⁵) It is also noted that thick intergranular and intragranular α can lead to crack blunting and crack growth deflection in the Ti-6.5Al-1.9Zr-0.25Si alloy.¹⁶)

It is well documented that the microstructure affects the fatigue properties of α + β titanium alloys greatly.¹⁵–¹⁸) The difference in FCGRs can be as high as 50-fold with the modification of microstructures in Ti-6Al-4V and Ti-6Al-6V-2Sn alloys.¹⁹,²⁰) In the open literatures, rather few works have been paid to investigate the mechanical properties of titanium welds, especially for Ti-6Al-6V-2Sn alloy. The present work focused on the notched tensile strength (NTS) and FCGR of Ti-6Al-6V-2Sn laser welds which were subjected to PWHT at various temperatures. The fracture appearance of a variety of specimens was examined with a scanning electron microscope (SEM). Additionally, mechanical properties of laser welds were correlated with the weld microstructures and fracture features.

2. Material and Experimental Procedures

Ti-6Al-6V-2Sn plates with a thickness of 4.0 mm were used in the mill-annealed condition, which had been heat-treated at 732 °C for 1 h, cooled in furnace to 482 °C, followed by air cooling to room temperature. The chemical composition of the alloy in weight percent was 5.50 Al, 5.40 V, 1.90 Sn, 0.5 Fe, 0.54 Cu, 0.02 C, 0.01 N, and the balanced Ti. The as-received material in the mill-annealed condition is named the base metal (BM) specimen, which consisted of a low percentage of β at the boundaries of elongated α grains in the microstructure. The hardness of Ti-6Al-6V-2Sn in the mill-annealed condition was approximately Hv 340. Tensile properties of the alloy included an ultimate tensile strength of 1117 MPa, a yield strength of 1086 MPa and an elongation of 20%.

A CO₂ laser was utilized for autogeneous bead-on-plate welding of the BM specimens in one pass with the welding direction normal to the rolling direction. A laser power of

*Corresponding author, E-mail: b0186@mail.ntou.edu.tw
2.7 kW and a scan rate of 800 mm/min were employed in the laser welding process. After welding, PWHTs were performed on the welds in the temperature range of 482 to 760°C for 3 h in vacuum, followed by argon-assisted cooling to room temperature. To distinguish the welds subjected to different PWHT temperatures, the numbers 482, 593, 704 and 760 were attached to the designated specimens. For instance, the W-482 specimen represented the weld which was subjected to a PWHT at 482°C.

Vickers microhardness measurements in distinct regions of the weld were conducted using a load of 300 g. The results were the average of at least three specimens for each testing condition. Fatigue crack growth tests according to ASTM E647-91 standard were conducted on CT specimens in a computerized MTS testing system. For laser-welded specimens, the crack growth was along the centerline of the fusion zone. In FCGR tests, the loading frequency was 20 Hz with a sinusoidal waveform at the stress ratio \( R \) of either 0.1 or 0.5 throughout the test. Crack length was monitored online using a crack opening displacement gage mounted at the specimen edge and confirmed by a 30X optical microscope.

The fracture appearance of various specimens was examined by a scanning electron microscope (SEM), with attention paid to the changes of fracture features. The detailed microstructures of the specimens were examined with a transmission electron microscope (TEM). A surface roughness tester coupled with software was used to determine the arithmetic average of the absolute values of the roughness profile, \( R_a \), and profile of fatigue-fractured specimens. The measurement of \( R_a \) was carried out at the central portion of the fatigue-fractured specimen along the crack growth direction over the \( \Delta K \) range of 9 to 12 MPa√m. Moreover, \( R_a \) could be used to index the extent of changing crack paths in the specimen.

3. Results and Discussion

3.1 Microhardness and microstructure

Figure 2 shows the variation of microhardness from the centerline of the WM to the BM in various welds. The low energy input of laser welding resulted in a narrow heat-affected zone (HAZ). As shown in Fig. 2, the WM is considerably harder than the HAZ and BM regions, regardless of the PWHT conditions. The peak hardness of the WM could reach Hv 460 after a PWHT at 482°C, i.e., the W-482 specimen. For the W-593 specimen, a decline in hardness to Hv 430 which was slightly lower than Hv 440 of the AW specimen was resulted. Moreover, an obvious decrease in the WM hardness was found for the welds which were post-weld heat treated at 704 or 760°C, in particular, the W-760 specimen. Nevertheless, the hardness in the WM was much higher than that in the BM, regardless of PWHTs.

![Fig. 2 Micro-hardness measurements in distinct regions of the welds.](image-url)
Figure 3 reveals TEM microstructures of the WM after various PWHTs. The AW specimen comprised of typically coarse columnar grains with fine needle features which were hardly resolved by optical microscope. TEM observations of the AW specimen revealed that fine acicular $\alpha'$ was distributed uniformly in the $\beta$ matrix (Fig. 2(a)). The absence of $\alpha'$-martensite in the specimen indicated that the cooling rate in laser welding was not fast enough to cause martensitic transformation. Fine acicular $\alpha'$ in the form of basket-weave resulted in the strengthening of the matrix and accounted for the high hardness (Hv 450) of the AW specimen. A slight increase in hardness of the W-482 specimen was attributed to the decomposition of retained $\alpha'$ during the PWHT. Moreover, thin layers of $\alpha'$ were found to decorate the grain boundaries of the W-482 specimen (Fig. 3(b)). In the case of the welds subjected to PWHTs at 593°C or higher, the coarsened microstructures led to the drop in hardness (Fig. 3(c)). Further increasing the PWHT temperature, e.g., 704°C, thick $\alpha'$ at the grain boundaries (Fig. 3(d)) was observed and expected to promote grain boundary deformation during tensile straining.

3.2 Notched tensile tests

Figure 4 shows the NTS values of the welds after PWHTs at different temperatures. The NTS of the BM specimen was about 1285 MPa, which was greater than 1117 MPa, the ultimate tensile strength of the BM specimen. It implied that notch-blunting of the BM specimen would cause a rise in strength before failure. Notably, the NTS of laser welds was lower than that of the BM specimen with the exception of the W-760 specimen. For example, the W-704 specimen had a NTS of 1100 MPa, approximately 185 MPa lower than the BM specimen, while the NTS of the AW specimen could only reach 940 MPa. It was noted that the welds with lower PWHT temperatures, e.g., 593°C and below, exhibited a remarkable drop in NTS (high notch brittleness) and the lowest NTS was associated with the W-593 specimen. Investigating the mechanical properties of Ti-6Al-6V-2Sn welds using gas-tungsten arc welding process, the weld metal possesses low toughness and ductility in the as-welded condition as compared to the base metal. To improve the weld properties, complex PWHT procedures are required for Ti-6Al-6V-2Sn welds, which might include the solution treatment at high temperature. Obviously, a single PWHT at the temperature at/above 704°C was useful to reduce notch brittleness of Ti-6Al-6V-2Sn laser welds.
3.3 Fatigue crack growth tests

The fatigue crack growth rates \( (da/dN) \) versus stress intensity factor range \( (\Delta K) \) curves for various specimens are shown in Fig. 5. The FCGRs of all welds are lower than those of the BM specimen within the experimental \( \Delta K \) range, regardless the stress ratio \( (R) \). Similar results have also been reported in Ti-6Al-4V laser welds.\(^8\)–\(^10\) It was deduced that the basket-weave microstructures in the WM had better resistance to crack growth than the elongated and banded structures in the mill-annealed BM. In general, the FCGRs of all specimens increased with increasing \( R \). It was also noticed that the W-482 specimen had a slightly lower FCGR than other specimens at \( R = 0.1 \) (Fig. 5(a)), but not the case at \( R = 0.5 \) (Fig. 5(b)). The reduced FCGR of the WM in the as-welded Ti-6Al-4V welds has been reported to be affected by the presence of residual stresses.\(^9\) The lower FCGR of the W-482 specimen at \( R = 0.1 \) could be attributed to only partial relief of welding residual stresses. On the other hand, the high notch brittleness of the W-482 specimen implied that monotonic fracture was more likely to occur at a higher \( R \), e.g., 0.5, therefore, enhanced the crack growth and exhibited a remarkable increase in FCGRs. Moreover, the W-760 specimen, which consisted of coarse structures in the matrix and thick \( \alpha \) at the columnar boundaries, had the lowest FCGR among the specimens at \( R = 0.5 \) (Fig. 5(b)).
3.4 Marco-fracture observations

Figure 6 shows the typically macroscopic fracture appearance of various notched tensile specimens. The BM specimen revealed obviously slant fracture regions in air (Fig. 6(a)). In the case of the welds with PWHTs at 704°C or higher, e.g., the W-760 specimen, the interior flat fracture and lateral slant fracture (Fig. 6(b)) on the fracture surface were observed. These indicated the weld possessed certain plasticity and underwent notch-blunting before rupture. In contrast, the fracture surfaces of the AW, W-482, and W-593 specimens, revealed extensively flat fracture regions (Figs. 6(c) and (d)). It was clear that refined $\alpha + \beta$ structures in the welds with PWHTs at 704°C or above, resulted in reduced notch brittleness. Furthermore, a flat intergranular fracture was also observed along coarse columnar grain boundaries of the W-593 specimen.

Figure 7 shows the macroscopic fatigue-fractured appearance of the BM and W-482 specimens tested at $R = 0.5$. The fracture surface morphology of the BM specimen was quite flat and revealed fine parallel strides in the direction of crack growth (Fig. 7(a)). Such characteristics could be due to the separations along the banded structure in the specimen, and became more obvious at high $R$. Besides, the fatigue-fractured appearance of all welds was alike, displaying irregular and coarse features of the solidification structures, independent of the stress ratio (Fig. 7(b)). A rougher fracture surface, which meant tortuous crack paths during crack propagation, was observed in the W-482 specimen compared to the BM specimen.

Figure 8 SEM fractographs of the weld with (a) crack growth normal to, (b) crack growth along solidification direction of the W-704 specimen, (c) transgranular fracture with grain boundary shear of W-760 specimen and (d) intergranular fracture along columnar boundaries of W-593 specimen after notched tensile tests.
growth, was expected to reduce the FCGR of the welds. Within the $\Delta K$ range of 9 to 12 MPa$\sqrt{m}$, the Ra values of the BM, W-482, W-593, W-704 and W-760 specimens at 0.1 $R$ were 2.35, 5.90, 7.73, 18.11 and 27.46 $\mu$m, respectively. The results also confirmed that the fracture surfaces of the welds were much rougher than the BM specimen, and increased with increasing the PWHT temperature. Fracture surface roughness is known to be an important parameter in controlling the fatigue crack propagation of the Ti-6Al-4V alloy.

3.5 Fractographic observations of notched tensile specimens

Figure 8 contains SEM fractographs showing the tensile fracture appearance of various specimens. The tensile fracture of the BM specimen exhibited extensive ductile dimple fractures. Because of the epitaxial growth and uneven dissipation of welding heat-input, columnar dendrites on the top were coarser than those on the bottom in a weld. Due to the lack of extensive slip systems in titanium alloys, coarse columnar grains tended to induce flat fractures in a macroscopic view. Transgranular fine dimples with grain boundary shearing were observed for the crack propagated normal to the solidification direction (Fig. 8(a)). However, the welds were more likely to form transgranular fracture with tear separations at interdendritic boundaries as the crack advanced mainly along the solidification direction (Fig. 8(b)). Such characteristics could be due to the presence of interdendritic $\alpha$ in the WM.

Transgranular dimple fractures across the columnar grains to-gether with grain boundary shearing (Fig. 8(c)) were observed in the W-760 specimen. It was evident that coarsened $\alpha + \beta$ structures with thick grain boundary $\alpha$
Improved the ductility and decreased notch brittleness. It was noticed that intergranular separations along coarse columnar boundaries was uncommon in the W-593 specimen (Fig. 8(d)). It was deduced that the extent of intergranular fractures along coarse columnar boundaries depended on the thickness of $\alpha$ at columnar grain boundaries, and the discrepancy in strength/hardness between the grain interior and boundary of the welds. The reason to cause the lowest NTS of the W-593 specimen among the welds is not known exactly. It was deduced the low NTS of the W-593 specimen partly be attributed to the premature fracture at the columnar grain boundaries, resulting in low NTS.

3.6 Fractographic observations of fatigue crack growth specimens

SEM fractographs revealing the fatigue fracture appearance of some specimens in various $\Delta K$ ranges and stress ratios are shown in Fig. 9. The fracture surface of the BM specimen showed typically transgranular fractures with fine ridges aligned in the crack growth direction (Fig. 9(a)) at $R = 0.1$. However, the BM specimen contained more fine dimples at higher $\Delta K$ or $R$ (Fig. 9(b)). These results could be partly attributed to the monotonic tensile fracture at high $\Delta K$ range or high $R$, likewise tensile fracture of the BM. Transgranular fatigue fractures with fine facets (Fig. 9(c)) were observed on all welds over the testing $\Delta K$ range. The presence of irregular facets on the fracture surface could be partly related to the limited slip systems of the alloy. At high $\Delta K$ or $R$, the welds also displayed a shear fracture at the columnar boundaries (Fig. 9(d)), in particular for the W-704 and W-760 specimens. The deflected crack growth at grain boundary $\alpha$ also led to a high Ra value in the welds. The improved resistance to crack growth for the welds after PWHT at 704°C or higher was attributed with ductile fractures of coarsened $\alpha + \beta$ structures (Fig. 9(e)) and the occurrence of grain boundary shearing (Fig. 9(d)) in comparison with other welds. At high $\Delta K$ range, secondary cracks were observed more frequently and associated with the interfaces cracking between $\alpha$ and $\beta$ phases (Fig. 9(f)).

4. Conclusions

Extremely fine acicular $\alpha$ distributed uniformly in the $\beta$ matrix of Ti-6Al-6V-2Sn laser welds could account for the high hardness of the AW and W-482 specimens. Ti-6Al-6V-2Sn laser welds exhibited notch brittleness or low NTS, unless a high PWHT temperature, e.g., 704°C, was employed. The tortuous crack path in the WM in contrast to the relatively straight path in the BM accounted for the lowered FCGR in the weld. The improved resistance to brittle fracture for the welds with PWHTs at 704°C or higher was attributed to the more ductile fracture of coarsened $\alpha + \beta$ structures in the matrix and the occurrence of grain boundary shearing. The fracture features of Ti-6Al-6V-2Sn laser welds were strongly affected by the inherent microstructures and crack growth direction with respect to the solidification or grain growth direction.

Acknowledgements

The authors gratefully acknowledge the financial support of this study by National Science Council of the Republic of China (NSC 96-2221-E-019-018).

REFERENCES