Interlaminar microstructure and mechanical properties of narrow gap laser welding of 40-mm-thick Ti-6Al-4V alloy

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Research Article

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Abstract

Narrow gap laser welding (NGLW) is a common solution for the welding of thick structures. NGLW was carried out on narrow-gap butt joints of 40-mm-thick Ti-6Al-4V alloy plates with U-shaped groove. The distribution characteristics of interlaminar microstructure in different height ranges of the joint were investigated, and the evolution behavior and formation mechanism of interlaminar microstructure of the joint were also clarified. It is found that a large number of short needle martensite nucleate and grow up near the fusion line and the upper boundary of the remelting zone. The “softening” phenomenon occurs in all welds except the cover layer weld. The microstructure evolution and defect migration induced by multiple welding thermal cycles in the upper weld forming process are the main reasons for the "softening" of the lower weld. The tensile strength of each sample changes in the range of 920 ~ 990 MPa and the fracture mode of the sample belongs to transgranular ductile fracture. In addition, compared with the upper part of the joint, the plasticity and toughness of the weld area in the lower part of the joint is improved.

1 Introduction

Titanium and its alloy, due to advantages, such as high specific strength, high corrosion resistance, low density, high toughness and fatigue resistance have been widely used in the field of aviation manufacturing. [1–3]. The most widely used Ti-6Al-4V alloy is an ideal material for manufacturing titanium alloy tubing, especially suitable for aviation hydraulic system and petroleum industry [4, 5]. Frame or beam components with a large size, unequal thickness, and variable cross section are important parts of fuselage structures of aerospace vehicles [6, 7]. Due to the large size, these structural components cannot be forged as a whole, so welding technology is particularly important. However, previous studies indicate that the existing manufacturing process of thick Ti-6Al-4V alloy parts has exposed several technical limits, such as suppression of welding defects and deformation controlling in the forming process [8, 9].

Currently, traditional arc-welding method still hold considerable importance for its advantages of cost control and convenient operation [10]. However, the arc-welding process existed in many weaknesses, such as instance excess distortions, tungsten inclusions [11]. Normally, the joint obtained by electron beam welding (EBW) is accompanied by deeper penetration and narrow heat-affected zone (HAZ). Nevertheless, the application of EBW is significantly limited by its vacuum condition [12, 13]. Narrow gap laser welding applies a laser to melt a filler material and fill a narrow U-shaped groove between two plates, which is a highly efficient process especially for welding thick metal plates [14, 15]. Many researches have reported that NGLW has superior ascendancy over other processes related to the welding of thick plates. According to Yu et al. [16], the technology of NGLW can overcome the penetration limit of the autogenous laser welding by layer-by-layer application of filler wire. Complementary studies by Sheriff et al. [17] demonstrated that the addition of filler material can compensate for the evaporation and burning-loss of elements during welding. As described in the study of Yang et al. [18], a defect-free welded
joint of 100 mm-thick SUS 304 steel plates are fabricated by NGLW with filler wire in the laser conduction mode.

As a result of the multiple welding processes, NGLW is accompanied by complex thermal effects. In addition to the remelting of the lower weld bead, the intense thermal effect could provide a driving force for the phase transition in the vicinity of fusion zone during the formation of the upper weld bead [19, 20]. According to the fundamental research on Ti-6Al-4V alloy joint from Zhang et al. [21], the proportion of primary α phase decreases and the proportion of transformed β phase increases with the variation of temperature gradient from near-base metal (BM) region to near-HAZ region. The study of Haghdadi et al. [22] has revealed the effect of thermal variation on crystallographic characteristics of α-α intervariant boundaries in Ti-6Al-4V alloy parts fabricated by additive manufacturing. Furthermore, sophisticated microstructure evolution will have implications for the service performance of the thick joint [23, 24]. The fatigue crack growth rates of laser powder bed fusion produced Ti-6Al-4V, which has been investigated by Becker et al. [25] and it is reported that the anisotropy of the fatigue properties is linked to morphological texture. Moreover, Maawad et al. [26] highlighted the tensile performance differs in various welded sheets having different volume fractions of grains with a certain orientation with respect to the welding direction. In conclusion, the studies on NGLW, there are few studies on the microstructure evolution of the interlayer in different regions of the thick joint in the welding process.

In the current investigation, 40 mm-thick Ti-6Al-4V alloy plates were welded successfully via the NGLW by fiber laser. The microstructure evolution of the interlayer in different regions of the thick joint was identified. The relationship between microstructure and mechanical properties of the joint was discussed in detail.

2 Experimental Details

2.1 Materials

The material of base metal and welding wire used in the narrow gap laser welding experiment are both Ti-6Al-4V alloy, of which the chemical composition and basic mechanical properties obtained by previous experiments are listed in Table 1 and Table 2, respectively. The 40 mm-thick Ti-6Al-4V alloy plates were machined with a double-sided U-groove, and its specific geometric dimension is presented in Fig. 1 (a) and (b).

<table>
<thead>
<tr>
<th>Composition</th>
<th>Fe</th>
<th>C</th>
<th>H</th>
<th>O</th>
<th>N</th>
<th>Al</th>
<th>V</th>
<th>Ti</th>
</tr>
</thead>
<tbody>
<tr>
<td>wt.%</td>
<td>≤ 0.30</td>
<td>≤ 0.10</td>
<td>≤ 0.015</td>
<td>≤ 0.20</td>
<td>≤ 0.05</td>
<td>5.5–6.8</td>
<td>3.5–4.5</td>
<td>Balance</td>
</tr>
</tbody>
</table>
### 2.2 Experimental setup

In the current work, the narrow gap laser welding experiment is performed by KUKA HR60HA robot and TruDisk-12003 Laser with the maximum output power of 12000 W produced by TRUMPF. The schematic diagram of NGLW is shown in Fig. 2, special designed wire feed nozzle with diameter of 3 mm was used and the molten pool was protected by high purity argon during laser welding process.

The whole welding process consisted of the formation of 21 layers of the weld bead. As presented in Fig. 1 (c), the scanning sequence of different weld bead inside the symmetrical U-shaped groove was alternating. In order to guarantee the satisfactory formation of the weld bead and sidewall fusion, a high defocusing was experimentally considered. The specific welding parameters of the NGLW process are given in Table 3.

### 2.3 Analysis method

Following the welding process, the metallographic sample was cut along cross-section of thick Ti-6Al-4V alloy plates, and uneven area of weld bead was avoided during the sample preparation. The appearance of macro and microscopic in welded joints are respectively observed by MR-5000 optical microscope (OM) after grinding with abrasive paper, polishing with diamond abrasion paste and etching with Kroll reagent (2% HF + 8% HNO$_3$ + 90% H$_2$O).

The Vickers microhardness measurements of different interlayers were conducted on a transverse cross-section of 40 mm-thick Ti-6Al-4V alloy joint with a load of 200 g and a residence time of 15 s. Static tensile testing an intuitive approach to reveal the mechanical behavior and basic performance indexes of metal materials such as yield strength, tensile strength, elongation at break when subjected to static load [27, 28]. In order to investigate the influence of interlayer microstructure on mechanical performance and fracture mechanism of 40 mm-thick Ti-6Al-4V alloy joint. The tensile experiments are performed with a
displacement rate of 0.8 mm/min at ambient temperature (25°C) using a universal testing machine (UTM). After tensile experiment, scanning electron microscopy (SEM) is adopted to observe the fracture morphology, while energy dispersive spectroscopy (EDS) is employed to characterize the element. Figure 3 marks the geometric dimension of test specimen and sampling location.

3 Result And Discussion

3.1 Weld formation

Figure 4 shows the macroscopic shape of the weld seam and the macroscopic shape of the joint cross-section, the cover layer weld is oxidized slightly, the weld surface is faint yellow. The filler material was well-spread in the groove and fully fused with the base material from both sides. There are only a few porosity defects in the joint cross-section, and the diameter of the porosity does not exceed 0.2 mm. The fusion between the layers and the side walls of the joint is well, and the fusion width of the weld is within the range of 7-7.5 mm, the height of the cover layer is about 3.5mm, in addition, the height of other filling welds is about 1.8-2.0mm.

Comparing the original model with the deformation data, the joint deformation cloud is depicted in Fig. 5, disregard of the displacement deviation caused by the unfilled position of the bevel and extracting the deformation data only for the edge of the test plate. The deformation results show that there is a slight warpage of the test plate after welding, and the maximum deformation of the test plate edge is only 0.25 mm, and the maximum warpage angle is calculated to be about 0.19°. Therefore, the symmetrical U-shaped groove was used and alternating welding sequence can suppress the deformation after welding, effectively.

3.2 Microstructure examination

The microstructure of the interlaminar weld zone is intercepted and studied in three different areas of the upper, middle and lower parts of the joint, respectively. The interlayer microstructure characteristics of different areas of the joint are analyzed, and the variability of the tissue morphology of the weld zone between different areas of the layers is quantitatively compared. Figure 6 demonstrates the interlayer microstructure of the cover layer. As seen in Fig. 6 (d) and (e), there are distinct upper boundary of the remelting zone and fusion line in the upper interlayer region of the joint. The measured statistics show that the width of the remelting zone is in the range of 900 ~ 1000 µm. Figure 6 (b) and (c) shows the microstructure near the interlayer fusion line, and it can be seen that from the figure that a short needle-like α' phase appears near the fusion line. Above it, i.e., the part of the cover layer weld, the widmanstatten structure consisting of long and wide clusters of needle-like martensite regularly arranged, among where the white color is the needle-like α' phase and the black border between the phases is the β phase; while below the fusion line, the basketweave structure consisting of cross-arrangement of needle-like martensite appears inside the coarse β columnar crystal. In addition, the second welding heat cycle in the
upper weld, the fusion line below the part of the original β columnar grain boundary gradually tends to be "blurred".

Figure 7 shows the interlayer microstructure morphology of the middle and lower part of the joint. Unlike the interlayer organization of the upper part of the joint, the upper boundary of the remelting zone in the middle and lower part disappears completely due to the occurrence of multiple solid-state phase transformations. Therefore, in this part, the interlayer microstructure near the fusion line of the two regions was intercepted for analysis. Figure 7 (c) and (d) shows that short needle-like martensite tissue also appears near the fusion line in the middle of the joint. As shown in Fig. 7 (d), the statistical measurements of the martensite length dimensions of ten groups in the field of view are randomly performed, and it can be seen that the average martensite length near the fusion line is around 12 µm. Figure 7 (f) and (g) show that although the width dimension of the columnar crystals in the lower part of the joint is smaller, the widmanstatten structure consisting of coarse needle-like martensite clusters still appears above the fusion line. In addition, as seen in Fig. 7 (g), the original β columnar grain boundary below the fusion line in the lower part of the joint has nearly disappeared.

The formation mechanism of the short needle-like α' phase near the upper boundary of the remelting zone and the fusion line is studied, as shown in Fig. 8. Ti-6Al-4V titanium alloy laser welding is a rapid heating and cooling process, in the process of rapid cooling of the high-temperature β phase, solid solution in which the alloying elements are too late to diffuse. At this time, the occurrence of martensitic phase transformation belongs to the non-diffusion type phase transformation. Many studies have shown that, similar to the liquid-solid phase transformation, martensite phase transformation is also a nucleation and growth process [8]. Therefore, this paper analyzed the formation mechanism of the short needle-like α' phase from the perspective of nucleation.

As shown in Fig. 8 (a), during the narrow gap laser welding of Ti-6Al-4V titanium alloy, the liquid molten pool can be divided into two parts: the lower weld, which is secondarily melted by the laser heat, and the new weld, which is formed by filling in the molten wire. While the formation of the upper weld, the surface area of the lower weld absorbs the laser energy and remelts. Subsequently, the wire melts and transitions into the molten pool under the action of the laser beam. During the transition of the molten droplet, a strong stirring effect on the formed molten pool is inevitable, as shown in Fig. 8 (b). Therefore, near the upper boundary of the remelting zone, there are obvious energy, structural as well as compositional undulations, which provide the nucleation conditions for the subsequent martensite phase transformation. At the same time, after solidification of the weld, a large number of defects such as interstitial atoms, dislocations, and laminations are retained near the boundary of the remelting zone due to the stirring behavior of the melt pool. When the high-temperature β phase cooled to the phase transition temperature, the high-density defects will provide a large number of nucleation conditions for the formation of the α' phase. Near the boundary, the α' phase first nucleates and grows at the location of point defects, line defects, etc. As a result, a large number of short needle-like α' phases are formed near the boundary of the remelting zone.
The formation mechanism of short needle-like α' phases near the fusion line is similar to the above-mentioned case. As shown in Fig. 8 (c), compared with the upper boundary of the remelting zone, the stirring behavior of the molten pool near the fusion line is relatively weakened. However, after liquid solidification, the original solid-liquid interface also retains a certain number of nucleation prone positions such as point defects, dislocations and stacking faults. In the meantime, because it is close to the base metal, the undercooling near the fusion line is higher. Many literatures show that when the undercooling increases, the critical nucleation decreases successfully, which means that the number of nucleation increases. Therefore, a large number of α' The phase nucleates and grows near the original solid-liquid interface, forming a short needle shape α'. The fusion zone formed by aggregation is shown in Fig. 8 (c).

Meanwhile, some of the β columnar crystals below the fusion line showed the fuzzy or even disappearance of grain boundaries, and this phenomenon appeared in several interlayer weld regions such as the upper, middle and lower parts of the joint, as shown in Fig. 9 (a) and (b). To explain the above phenomenon, this study is carried out to investigate the microstructure evolution behavior of the heat-affected zone below the fusion line. As seen in Fig. 9 (a) and (b), the grain boundary disappearance region is basically located within 200 µm below the fusion line, and this region is subject to the same strong thermal action during the forming of the upper layer weld, although remelting does not occur. Due to the different cooling conditions, the microstructure before and after the solid-state phase transformation may differ; at this moment the original β columnar grain boundary morphology is weakened.

In addition, the grain boundary is a collection of a large number of point defects and line defects (dislocations). First, the local area below the fusion line experiences a sudden temperature rise and increased atomic activity due to the thermal influence of the upper weld. As a result, point defects concentrated at grain boundaries start to migrate under the action of driving force, and some vacancy defects migrate to dislocations or disappear in migration by combining with interstitial atoms. At the same time, due to the increased atomic activity, in addition to point defects, dislocations in the role of internal stresses in the weld is activated and slip, while the slip process of dissimilar dislocations occur in combination and offset. Therefore, the grain boundary near the point defects, line defects density will be further reduced. In summary, due to the tissue evolution of the secondary solid state phase transformation and the influence of defect migration, the fusion line below part of the original β columnar grain boundaries in the upper layer of the weld forming process under the influence of heat gradually tends to blur or even disappear.

### 3.3 The weld joint mechanical properties

Figure 10 shows the path selection schematic for the micro-Vickers hardness test, and a total of four paths were selected. The hardness test results are shown in Fig. 11. The results show that the joint of heat-affected zone and weld hardness are much higher than the base material after the martensitic phase transformation is completed. The weld area below the cover layer shows an obvious "softening"
phenomenon. The average hardness of the cover layer weld is 379.4 HV, while the average hardness level of the other layers of the weld is only 339.1 HV. Analysis shows that the upper layer of the weld forming process of multiple welding thermal cycle-induced tissue evolution and defect migration behavior is the main reason for the "softening" of the lower layer of the weld.

In order to further investigate the influence of interlayer tissue variability in different areas of the joint on the mechanical properties, the microstructure of specimen #5 was analyzed. The failure of the specimens occurred in the parent material area, and the microscopic morphology near the fracture surface of specimen #5 showing in Fig. 12. The fracture surface penetrates a large number of equiaxed grains in the base material area. Therefore, the fracture mode of the tensile specimens can be tentatively judged as grain penetration ductile fracture.

Figure 13 (a) shows the displacement-stress variation curves of different tensile specimens. During the tensile process, with the displacement is increased, the specimen first enters the elastic deformation stage in where the stress level increases linearly. Subsequently, when the specimen deformation exceeds the critical value, the tensile specimen enters the plastic deformation stage in where the deformation is irreversible. Finally, with the increasing loading force, the microcracks inside the specimen continued to expand to the critical length and the specimen failed to fracture. Comparing the test results of tensile specimens sampled from different areas of the joint, it can be seen that the tensile strength of the specimens ranged from 920 to 990 MPa, and the tensile strength of different specimens did not very much, and did not show an obvious pattern of change.

The post-break length L and post-break cross-sectional area A were measured for the failed specimens, and the results are shown in Fig. 13 (b). The measurement results showed that the specimen #1 had the highest shrinkage at break, reaching 20%, while the specimen #5, which was sampled from the upper part of the joint, had a shrinkage at break of only 8.96% as the sampling position increased. Similarly, the elongation after break showed the same trend, with specimens #1 and #2, which were sampled from the lower part of the joint, showing an elongation after break of about 4%, while the elongation after break of specimen #5 decreased to 2.55%. Tensile specimens can be divided into three parts: base material, heat-affected zone and weld zone, and all specimens have the same mechanical properties base material, so different specimens in the tensile test show plasticity variability associated with the weld, heat-affected zone.

The fracture morphology observation of the tensile specimen was carried out, as shown in Fig. 14. From the figure, the fractures are fibrous and show a grayish color. The fracture areas are all equiaxed tough nests, which can be judged that the fracture modes of the tensile specimens all belong to the microporous aggregation type of toughness fracture. In addition, the composition scan of the local area of the fracture shows that the mass fraction of V element in the fracture section has increased, while the content of Al element has decreased.

4 Conclusions
The microstructure evolution of the interlayer in different regions of the thick joint and the relationship between microstructure and mechanical properties of Ti-6Al-4V alloy manufactured by the narrow gap laser welding by fiber laser have been investigated in detail. The following specific conclusions can be drawn:

1. There is still a clear upper boundary of the remelting zone at the lower part of the cover layer weld. Except for the cover layer weld, the upper boundary of the remelting zone between other weld layers disappears completely after multiple welding thermal cycles. A large number of short needles are formed near the boundary of remelting zone and fusion line α’ phase.
2. The hardness test results show that there is an obvious "softening" phenomenon in the weld area below the cover layer. The microstructure evolution and defect migration induced by multiple welding thermal cycles in the upper weld forming process are the main reasons for the "softening" of the lower weld.
3. The tensile property test results show that the tensile strength of each sample changes in the range of 920 ~ 990 MPa. The fracture surface and fracture characteristics show that the fracture mode of the sample belongs to transgranular ductile fracture. In addition, the plastic toughness of the weld zone in the lower part of the joint is improved significantly compared with that in the upper part.

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**Ethics approval**

Not applicable.

**Consent to participate**

Not applicable.

**Consent for publication**

All authors have read and agreed to the published version of the manuscript.

**Conflict of interest**

The authors declare no competing interests.

**Availability of data and material**

Data and materials are available.
Author contribution

Xing Liu: Investigation, Writing – original draft, Writing – review & editing. Wanli Ling: Methodology, Writing – review & editing. Yue Li: Investigation. Jianfeng Wang: Data curation, Validation. Xiaohong Zhan: Supervision, Project administration, Writing – review & editing.

References


28. Y. Zhang, Y. Bi, J. Zhou, D. Sun, H. Li, Microstructure and mechanical property improvement of Ti alloy and stainless steel joint based on a hybrid connection mechanism. Materials Letters, 274 128012 (2020)

**Figures**

![Figure 1](image)

**Figure 1**

The schematic diagram of weldment: (a) and (b) geometric dimension of 40 mm-thick Ti-6Al-4V alloy plates, (c) welding sequence
Figure 2

(a) and (c) the schematic diagram of NGLW process, (b) NGLW experimental equipment
Figure 3

Tensile test specimens: (a) and (b) sampling location, (c) sample dimensions

Figure 4

Macro forming of 40 mm thick TC4 titanium alloy narrow gap laser welded joint: (a) macro forming of welded joint, (b) cover layer, (c) joint cross-section
Figure 5

Deformation cloud after welding of 40 mm thick TC4 titanium alloy narrow gap laser welded joint
Figure 6

Microstructure of the upper interlayer weld area of the joint: (a) joint cross-section, (b) and (c) microstructure near the fusion line, (d) and (e) microstructure near the upper boundary of the remelting zone.
Figure 7

Microstructure of the middle and lower interlayer weld area of the joint: (a) macromorphology, (b) and (c) interlayer microstructure in the middle of joint, (d) and (g) needle-like $\alpha'$ phase morphology, (e) and (f) interlayer microstructure near the root of joint.
Figure 8

Analysis of the mechanism of microstructure evolution in the interlayer weld area: (a) diagram of weld zone, (b) microstructure evolution near upper boundary of remelting zone, (c) microstructure evolution of fusion zone.
Figure 9

Analysis of the mechanism of the influence of welding thermal cycle characteristics on the heat-affected zone below the fusion line: (a) cover layer, (b) intermediate layer, (c) microstructure evolution of original β grain boundary during welding, (d) microstructure of original β grain boundary after welding
Figure 10

Vickers hardness test path selection diagram

Figure 11

Vickers hardness test results of welded joints: (a) path 1 - path 3, (b) path 4
Figure 12

Macroscopic morphology and microstructure of tensile specimen cross-section at failure position: (a) schematic diagram of metallographic sampling location, (b) macroscopic appearance of cross section of failure location, (c) microstructure morphology
Figure 13

Tensile property test results: (a) displacement-stress variation curve of tensile specimens, (b) elongation at break and shrinkage at section of tensile specimens

Figure 14
Fracture microscopic morphology and composition of tensile specimens: (a)-(c) the SEM results, (d) the EDS result