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Microstructures and Mechanical Properties of a Laser-Welded Joint of Ti₃Al-Nb Alloy Using Pure Nb Filler Metal

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Received: 9 September 2018; Accepted: 29 September 2018; Published: 1 October 2018

Abstract: Ti₃Al-Nb alloy (Ti-24Al-15Nb) was welded by a pulsed laser welding system without and with pure Nb filler metal. The results indicated that pure Nb filler metal had profound effects on the microstructures and mechanical properties of the laser-welded joints. The joint without filler metal consisted of the weld zone ($\alpha'_2 + B2$), heat affected zone HAZ1 ($\alpha_2 + B2$), HAZ2 ($\alpha_2 + O + B2$) and base metal ($\alpha_2 + O + B2$), and gas pores were generated in the weld resulting in the deterioration of the joint strength (330 MPa) and elongation (1.9%). When the Nb filler metal was used, the weld microstructure (NbTi solid solution + O + B2) was obtained, and the joint properties were significantly improved, which was associated with the strengthening effect of the NbTi solid solution, O phase precipitation and the slip transmission between O and B2 phases, and the restraining of the formation of martensite (α'_2) and gas pores in the weld. The strength (724 MPa) and elongation (5.1%) of the joint increased by 119.4% and 168.4% compared with those of the joint without filler metal, and the joint strength was able to reach 81.7% of the base metal strength (886 MPa). It is favorable to use pure Nb filler metal for improving the mechanical properties of laser-welded Ti₃Al-Nb alloy joints.

Keywords: Ti₃Al-Nb alloy; laser welding; pure Nb filler metal; microstructures; properties

1. Introduction

Ti₃Al-based alloys are considered as ideal elevated temperature structural materials with light weight in aircraft and aerospace fields, due to their low density, high specific strength, and excellent resistance to oxidation and creep. In addition, Ti₃Al-based alloys have a higher service temperature than titanium alloys and can be also 40% lighter than nickel-based superalloys [1,2]. However, because of the low room temperature ductility of these alloys, it is difficult to process and machine them at room temperature, and hence their practical application is limited for a period of time. In recent years, the ductility and fracture toughness of Ti₃Al-based alloys have been greatly improved by alloying with metals such as Nb, V, Mo, etc. [3]. Thus, their application has been awaited. It is well known that joining technology is one of the most important methods in materials processing. The successful joining of Ti₃Al-based alloys would be favorable for the broadening of their application fields. There have been some research activities on the joining of Ti₃Al-based alloys. The joining methods used include diffusion bonding [4], friction welding [5], brazing [6], tungsten inert-gas arc welding (TIG), electron beam welding (EBW) and laser beam welding (LBW), and so on. However, the application of solid-state welding and brazing is limited by the geometry and size of components. Fusion welding is more convenient in welding large-sized and complex components [7–11]. At present, the research work on the fusion welding of Ti₃Al-based alloys mainly focuses on its weldability, microstructure and on improving the performance of welded joints [12–20].

Liu studied the TIG weldability of Ti-23Al-14Nb-3V alloys [15]. The result indicated that preheating had an obvious effect on the weldability. The TIG-welded joint with cold cracking exhibited poor strength (245.80 MPa) under a condition without preheating. When suitable preheating was applied, the cold cracking could be dispelled and joint strength was increased (637.95 MPa). Chen reported the microstructures and tensile properties of Ti-22Al-25Nb EBW joints. The as-welded fusion zone had a fully B2 microstructure, and the strength and elongation of the joint could be improved after post-weld annealing [16]. The research of David et al. indicated that welding speed had a profound effect on the microstructure and hardness of EBW joints of Ti-14Al-21Nb (wt %) [17]. The fusion zone microstructure contained a predominantly fine, acicularly ordered α_2 phase at low welding speeds and a predominantly metastable B2 phase at high welding speeds. Microhardness profiles across the joints were more uniform at low welding speeds than at high welding speeds. Wu studied the effects of laser welding parameters on Ti-24Al-17Nb joints [18]. The results indicated that the laser weld microstructure consisted primarily of a retained ordered β phase (namely B2 phase) and was independent of the laser welding parameters, while the joint bend ductility was related to the welding conditions. Wang et al. investigated the bending properties of laser-welded Ti-23Al-17Nb alloys, finding that the joint bending ductility decreased as the laser heat input increased [19]. Cieslak described the effect of cooling rate on the microstructure of Ti-26Al-11Nb joints, showing that the particular microstructure produced was primarily a function of the cooling rate from the β solvus, and the rapid cooling rate of laser welding resulted in a microstructure in which the high temperature β underwent an ordering reaction on cooling to the B2 crystal structure [20]. From the viewpoint of welding metallurgy, the filler metal should have an effect on the microstructure and mechanical properties of welded joints of Ti₃Al-based alloys. However, up to now, the research work dealing with this aspect has been very limited.

The present work investigates the effects of pure Nb filler metal on laser-welded joints of Ti₃Al-Nb alloys (Ti-24Al-15Nb). Nb was selected as a filler metal based on Nb and Ti having ultimate mutual solubility. The relationship between joint microstructure and mechanical properties was studied and discussed. The purpose of this is to better understand the weldability of Ti₃Al-Nb alloys and to provide the foundation for improving the mechanical properties of the laser welded joints.

2. Experimental Procedure

In this experiment, a Ti₃Al-Nb alloy plate with a composition of Ti-24Al-15Nb (Table 1) and size of 50 mm × 15 mm × 1 mm was used as a base metal, and a pure Nb sheet with a purity of 99.99% and thickness of 0.3 mm was selected as the filler metal. Figure 1 shows the microstructures of the base metal consisting of α_2 , O and B2 phases, with the granular α_2 and lath-shaped O phases distributed within B2 matrix. The mechanical properties of the base metal are presented in Table 2.

Figure 1. Microstructures of the base metal: (a) optics microscopy (OM) image and (b) SEM image.

Table 1. Chemical compositions of the Ti-24Al-15Nb alloy base metal.

Elements	Ti		Al		Nb	
	wt %	at %	wt %	at %	wt %	at %
Base metal	58.14	60.27	13.26	24.43	28.60	15.30

Table 2. Mechanical properties of Ti-24Al-15Nb alloy at ambient temperature.

Tensile Strength (MPa)	Yield Strength (MPa)	Elongation (%)	Micro-Hardness (HV)
886	620	12.8	365

Prior to welding, Ti₃Al-Nb alloy plates and the Nb sheet were ground, then placed in a specific solution of 3% HF + 30% HNO₃ + 67% H₂O for 40 s, and finally ultrasonically washed with acetone in order to remove oxides and dirt. The Ti₃Al-Nb/Nb/Ti₃Al-Nb (Ti₃Al-Nb/Ti₃Al-Nb) assembly was kept in contact in a self-made fixture, and the welds were made with Nb filler metal and without filler metal using a Nd:YAG pulsed laser welding system (HKW-1050B) with a wavelength of 1.064 μ m, as shown in Figure 2. According to a previous study [21], a laser power of 222 W, a pulse current of 88 A, pulse duration of 10 ms, pulse frequency of 6 Hz, laser beam diameter of 0.1 mm, focal length of 150 mm, welding speed of 200 mm/min and Ar shielding gas flow of 18 L/min were used in this experiment. The focus position was on the surface of the two plates.

Figure 2. Schematic diagrams of the laser-welded Ti₃Al-Nb alloy (a) without filler metal and (b) with pure Nb filler metal.

After welding, the metallographic specimens of the laser-welded joints without and with Nb filler metal were cut vertical to the weld, and then they were ground, polished and etched with Kroll's reagent. The microstructure of the laser-welded joints was examined using optics microscopy (OM, ZEISS Scope A1, Jena, Germany), scanning electron microscopy (SEM, VEGA3, Jena, Germany), transmission electron microscopy (TEM, JEM-2100F, Tokyo, Japan) and micro X-ray diffraction (M-XRD, D8 Discover with GADDS, Bruker, Karlsruhe, Germany). The TEM sample was prepared by three steps: (1) a sample with the size of 10 mm $\times$ 10 mm $\times$ 0.3 mm was cut down from the laser-welded joint by an electro-discharge machine, and then was ground to 30 μ m in thickness; (2) it was corroded slightly to distinguish the weld zone, HAZ and base metal, and punched in the middle of the weld zone by a thin needle to form a sample with a diameter of 3 mm and thickness of 30 μ m; (3) the sample was ion-thinned with a gun-to-sample angle of 15° to form the sample for TEM observations. The specific experimental details of XRD measurements were a working tube voltage of 40 kV, a tube current of 40 mA, a scanning speed of 5 degrees/min and a scanning time of 14 min. The joint composition was analyzed with an energy dispersive X-ray spectroscopy (EDS, Oxford Aztec, London,

England). The mechanical characterization of the joints was evaluated by hardness and tensile tests. The microhardness across the joints was measured using a microhardness tester (MH-3, Xi'an, China) with a load of 200 g and dwell time of 10 s. The static tensile test was carried out at room temperature using a material testing system (MTS810) at a tensile rate of 0.5 mm/min. The location and size of the tensile specimens are shown in Figure 3. The tensile strength and elongation of the joints were determined based on the average value of three measurements per condition. After the tensile test, the joint fractured surface was examined using SEM.

Figure 3. Schematic of the laser-welded joint (a) and size and shape of the tensile specimen (b).

3. Results and Discussion

3.1. Microstructures of Laser-Welded Joints

The appearance of a weld is one of the most important factors when evaluating welding quality. Figure 4 shows the face and back morphology of laser-welded $\text{Ti}_3\text{Al-Nb}$ alloy joints without filler metal and with pure Nb filler metal. As can be seen, both laser-welded joints exhibit a good weld appearance, and there are no welding defects such as surface cracks and incomplete penetration. The experimental result indicated that the laser-welded joints consist of weld zone (WZ), heat affected zone (HAZ) and base metal (BM). The joint microstructure was studied using OM, SEM, TEM, XRD and EDS.

Figure 5a shows the weld morphology of the laser-welded $\text{Ti}_3\text{Al-Nb}$ alloy joint without filler metal. The semi-circular laminar lines can be seen at the top and bottom of weld; in addition, in the middle and the lower part of the weld, the laminar line is almost parallel to the fusion line, which is mainly related to the heat transfer of the weld pool and flowing characteristics of the liquid metal [21,22]. At the initial stage of laser welding, the heat transfer was mainly carried out along the cross-direction and depth direction; thus, the semi-circular shape was formed at the upper part of the weld. Then, the heat was mainly transferred along the depth direction to increase the depth of penetration [23], thus producing a laminar line parallel to the fusion line. In addition, the periodic changes of heat input and the release of latent heat of crystallization resulted in forming multiple laminar lines parallel to the fusion line during the welding process [24,25]. When the heat was transferred to the bottom of the weld pool, the direction of heat transfer was transformed from the longitudinal direction to the cross direction, thus forming a semi-circular shape at bottom of the weld, which increased the weld width at the bottom of the weld.

From Figure 5a, some gas pores were observed in the weld zone. Unlike that of arc welding, the gas-pore formation of laser welding may be due to the trapping of metal vapor because of the higher power density used for laser welding [26–28]. EDS analysis indicated that the weld Al content (8.00 wt %) was reduced compared with the base metal (13.26 wt %), and the Al content (9.62 wt %) of gas pore wall was higher than that of weld, as shown in Figure 5b and Tables 1 and 3. These results suggested that the weld pores should be related to Al vapor. During laser welding, Al was vaporized to form bubbles in the weld pool, because it had a higher vapor pressure and lower boiling point than Ti and Nb. The Al vapor escaped to the weld pool surface, resulting in its level reducing. The weld

Al content and the trapped Al bubbles formed gas pores in the weld as the temperature dropped and the pool solidified. In addition, the weld pores are also related to the fast cooling rate for laser welding. Due to the large temperature gradient for laser welding, the weld presents a fine columnar grain structure (Figure 5c). The columnar grain tended to grow in a direction perpendicular to the fusion line because this is the direction of the maximum temperature gradient and hence of maximum heat extraction. The B2 phase was detected in the weld zone by XRD analysis, as shown in Figure 5d. TEM was used to reveal more detailed microstructural characteristics of the weld zone, and the results are shown in Figure 6. Apart from the B2 phase (Figure 6a,b), needle-like martensite (α'_2) was found in the weld zone (Figure 6c). According to the Ti-Al alloy phase diagram, during the solidification of the weld pool, the β phase was precipitated first from the pool ($L \rightarrow \beta$). With the temperature dropping, the β phase transformed into the metastable B2 phase ($\beta \rightarrow B2$), and this could be also transformed into the martensite (α'_2) under suitable cooling conditions [17]. Based on the results above, for the weld produced without filler metal, the gas pores mainly resulted from Al vapor and the needle-like martensite (α'_2) was precipitated within the B2 matrix, which is harmful for mechanical properties.

Figure 4. The face and back morphology of laser-welded joints: (a) face of weld without filler metal; (b) back of weld without filler metal; (c) face of weld with Nb filler metal; and (d) back of weld with Nb filler metal.

Figure 5. Cont.

Figure 5. The weld morphology (a), EDS result (b), weld microstructure (c) and XRD result (d) of laser welded joint without filler metal. Base metal (BM) and weld zone (WZ) in Figure 5b.

Figure 6. TEM results of weld zone without filler metal: (a) B2 phase, (b) selected area diffraction pattern, and (c) needle-like martensite.

Table 3. The chemical composition of the weld zone and gas pore wall in the weld without filler metal.

Elements	Ti		Al		Nb	
	wt %	at %	wt %	at %	wt %	at %
Weld zone	63.09	68.44	8.00	15.40	28.91	16.16
Gas pore wall	62.05	66.17	9.62	18.23	28.32	15.60

In order to improve the microstructures and properties of the weld without filler metal, the effect of pure Nb filler metal was studied. Figure 7 shows the weld morphology, weld microstructure, EDS and XRD results of a laser-welded joint produced using pure Nb filler metal. From Figure 7a, it can be seen that the weld zone is non-uniform, consisting mainly of both a white region and gray region. This can be explained by the incomplete mixing of melted Nb filler metal and the partially melted base metal in the weld pool. In addition, the gas pores were not observed in the weld. Due to the addition of Nb filler metal and the reduction of the fusion ratio of the base metal, the weld Nb

content increased obviously and its Ti and Al contents decreased compared with base metal (Figure 7b). The EDS analysis indicated that the distribution of Ti, Al and Nb in the weld is heterogeneous (Table 4). The white region in the weld had higher Nb content (72.24 wt %) than the gray region (56.02 wt %), and hence it was the NbTi solid solution. Figure 7c corresponds to a magnification image of the weld zone shown in Figure 7a. As can be seen, the white region (NbTi solid solution) displays an island structure and the gray region contains columnar crystal and dendrite. The columnar crystal nucleated from the liquid metal upon the NbTi solid solution grains and then transformed into the dendrite with the advance of the solid/liquid interface. The results indicated that the NbTi solid solution was precipitated first from the weld pool.

Figure 7. The weld morphology (a), EDS result (b), weld microstructure (c) and XRD result (d) of a laser-welded joint produced with pure Nb filler metal.

Table 4. The composition of the white region and gray region in the weld with Nb filler metal.

Elements	Ti		Al		Nb	
	wt %	at %	wt %	at %	wt %	at %
Gray region	36.28	45.99	7.70	17.35	56.02	36.66
White region	23.78	34.90	3.98	10.38	72.24	54.72

Due to the higher melting point of Nb (2468 °C), it could act as the substrate for nucleation. The solidification mode changing from columnar crystal to dendrite is associated with the degree of the constitutional supercooling continuing to increase. According to the XRD analysis result (Figure 7d), the weld zone was mainly composed of NbTi solid solution, O and B2 phases. The weld was also examined using TEM, and its bright field image and selected area diffraction patterns are given in Figure 8. From the TEM results, the lath-shaped O phase precipitating within B2 matrix was identified, and the martensite (α'_2) was not found in the weld.

Figure 8. TEM results of a weld zone with Nb filler metal: (a) bright field image; (b) selected area diffraction pattern of B2 phase and (c) selected area diffraction pattern of O phase.

The microstructural characteristics of the weld depend on the chemical composition changing. According to Tables 3 and 4, the weld with Nb filler metal has higher Nb content (56.02–72.24 wt %) and lower Al content (3.98–7.70 wt %) than the weld without filler metal (28.91 wt % Nb and 8.00 wt % Al). Nb is a stabilizing element of B2 [3], promoting the formation of a B2 phase in the weld with increasing Nb content. With the further increase of Nb content, the upper temperature limit of the forming O phase increased and the alloy had a larger phase region of O + B2 phases [29–31]; hence, the O phase precipitated in the weld rich in Nb. Nb has an effect on the martensite transformation temperature and it reduces with increasing Nb content [25]. The martensite disappearing in the weld rich in Nb can be explained by its transformation temperature being well-below room temperature. In addition, because Al is a stabilizing element of martensite [20], the decreased weld Al content also inhibited the formation of martensite. It should be pointed out that the disappearance of gas pores in the weld is related to the decreased weld Al content. The smaller the weld Al content is, the less Al vapor there is, and hence the less Al pores there are in the weld.

Figure 9a,b show the morphology of the heat-affected zones (HAZs) in the laser-welded joints without and with Nb filler metal, respectively. The width of both HAZs ranged from 400 μm to 470 μm. There was a transition of microstructures from the fusion line towards the base metal, which is mainly attributed to experiencing different welding thermal cycles during laser welding. As the distance from the fusion line increased, the peak temperature and cooling rate of welding thermal cycles dropped, resulting in the formation of different microstructures. According to microstructural characteristics, the HAZs can be divided into HAZ1 and HAZ2. The HAZ1 is just adjacent to the fusion line and consists mainly of the granular α_2 and B2 matrix. The α_2 has a coarsening tendency compared with that in the original base metal, as shown in Figure 9c. During laser welding, the HAZ1 experienced a higher peak temperature than β /B2 transition temperature. At the heating stage, the O and B2 in the original base metal were transformed into β phase ($O + B2 \rightarrow \beta$). Because of the fast heating rate for laser welding and a relatively small rate of the $\alpha_2 \rightarrow \beta$ phase transformation, the α_2 was partially transformed into β , and the retained α_2 coarsening occurred. At the cooling stage, the β

phase was transformed into B2 phase and the retained α_2 phase remained at the ambient temperature, forming the microstructures of coarse α_2 +B2. It should be pointed out that the $\beta \rightarrow \alpha'_2$ martensite phase transformation could also occur under suitable cooling conditions [12]. The HAZ2 is close to the base metal and consists mainly of α_2 , O and B2 phases, as shown in Figure 9d. The peak temperature of HAZ2 was lower than the β /B2 transition temperature and it dropped as the distance from the fusion line increased. During laser welding, the O and α_2 were gradually transformed into B2 phase, and the $O + \alpha_2 \rightarrow B2$ partial phase transformation occurred with a dropping of peak temperature. The lower the peak temperature was, the more the O and α_2 phases were retained. Due to the fact that the cooling rate in HAZ2 is relatively lower than that in HAZ1, the fine lath-like O phase was precipitated from the B2 matrix [32], and the retained O and α_2 remained at the ambient temperature, hence forming the microstructures of $\alpha_2 + O + B2$ in the HAZ2.

Figure 9. Microstructures of heat-affected zones (HAZs): (a) HAZ without filler metal; (b) HAZ with Nb filler metal; (c) microstructure of HAZ1 and (d) microstructure of HAZ2.

3.2. Mechanical Properties of Laser-Welded Joints

The experimental results indicated that the Nb filler metal had profound effects on the mechanical properties of laser-welded Ti₃Al-Nb alloy joints. Figure 10 shows the microhardness profiles across the laser-welded joints without and with pure Nb filler metal. The microhardness distribution of the welded joints was not uniform, and the Nb filler metal had an obvious effect on the weld microhardness. For the joint without filler metal, the highest microhardness appeared in the weld zone (WZ), and its average hardness was 391 HV, followed by HAZ (375 HV) and BM (366 HV). Based on the study above (Figures 5 and 6), the WZ consists mainly of α'_2 and B2. The significant enhancement of WZ hardness is mainly attributed to forming a hard needle-like martensite (α'_2) within the B2 matrix. For the joint with Nb filler metal, the WZ microhardness (350 HV) was lower than HAZ and BM. The reduced weld microhardness is mainly associated with the increased Nb content restraining the martensite phase transformation and precipitating the NbTi solid solution. The microhardness of HAZs has an increased tendency compared with that of BM, which should be related to precipitating the fine lath-like O phase and martensite in the HAZs [12,33].

Figure 10. Microhardness profiles of laser-welded joints without and with Nb filler metal.

Figure 11 illustrates the stress-strain curves of $\text{Ti}_3\text{Al-Nb}$ alloy (BM) and laser-welded joints without and with Nb filler metal. The average tensile strength and elongation of BM were 886 MPa and 12.8%, respectively. The joint without filler metal showed poor tensile strength (330 MPa) and elongation (1.9%), as shown in Figure 11a, only accounting for 37.2% and 14.8% of BM strength and elongation. When the Nb filler metal was used, the joint properties were significantly improved (Figure 11b). The strength (724 MPa) and elongation (5.1%) of the joint increased by 119.4% and 168.4% compared with those of the joint without filler metal. The joint strength reached 81.7% of BM strength (886 MPa).

Figure 11. Stress-strain curves of the base metal (BM) and laser-welded joints without filler metal (a) and with Nb filler metal (b).

In order to clarify the reasons for the joint properties changing, the fracture characteristics of the laser-welded joints under tensile stress were studied. Figure 12 shows the fracture path and fracture surface morphology of the joint without filler metal. From Figure 12a, it can be seen that the joint fracture occurred in the weld zone. Its fracture surface displayed river-like patterns with brittle cleavage surfaces, and the gas pore was observed on the fracture surface, as shown in Figure 12b. The results demonstrated that a weld with a high hardness and brittleness is the weakest zone for the joint without filler metal. The reasons for these poor joint properties can be attributed to the embrittlement of the weld ($\alpha'_2 + \text{B}_2$) and the formation of gas pores. The former facilitated the initiation and propagation of cracks under tensile stress and the latter not only provided preferential sites for the initiation and propagation of cracks, but also decreased the effective cross-sectional area of the weld.

Figure 12. The fracture path (a) and fracture surface morphology (b) of the joint without filler metal.

Figure 13 shows the fracture path and fracture surface morphology of the joint with Nb filler metal, which is different from those shown in Figure 12. The crack was initiated at the weld toe due to generating stress concentration under tensile stress and propagated in HAZ1, as shown in Figure 13a. The fracture surface also had some cleavage fracture features, but fine dimples and tear-shaped marks along the dimple boundaries were observed (Figure 13b). Therefore, the significant enhancement of the joint properties is mainly attributed to the Nb filler metal improving the weld microstructure. From Figures 7 and 8, the weld zone consists mainly of a white region (NbTi solid solution) and gray region (O + B2), without weld defects such as gas pores. The NbTi solid solution has a body-centered cubic (bcc) crystal structure, favoring the improvement of joint properties. According to references [16] and [34], the closest-packed planes (001) in the O phase grow parallel to the close-packed planes (110) in the B2 phase, and the orientation relationship (OR) between these two phases are $\langle 111 \rangle$ B2 // $\langle 110 \rangle$ O. Hence the $\langle 111 \rangle$ (110) B2 slip is considered to be readily transmitted to [110] (001) O slip and the OR is compatible with the active slip systems in the O and B2 phases. Such slip transmission is thought to reduce the stress concentration effects at the grain boundaries of O and B2. Hence, the dual phase structure of B2 and O phases can improve the tensile strength and elongation of weld metal. In addition, the increased weld Nb content and decreased Al content inhibited the formation of martensite and gas pore in the weld, which also favor the improvement of joint properties. Therefore, it is favorable to weld Ti₃Al-Nb alloy using pure Nb filler metal for improving the mechanical properties of welded joints.

Figure 13. The fracture path (a) and fracture surface morphology (b) of joints with Nb filler metal.

4. Conclusions

1. The laser-welded Ti₃Al-Nb alloy joint without filler metal consists of WZ (α'_2 + B2), HAZ1 (α_2 + B2), HAZ2 (α_2 + O + B2) and BM (α_2 + O + B2). The joint fracture occurs in the WZ with higher hardness under tensile stress, and its strength and elongation are 330 MPa and 1.9%, respectively.

The deteriorated joint properties are associated with the embrittlement of WZ ($\alpha'_2 + B2$) and formation of gas pores.

2. The pure Nb filler metal has profound effects on the composition and microstructure of the laser weld. The weld Nb content increases and Al content decreases compared with those of welds without filler metal. The weld microstructure is characterized by NbTi solid solution + O + B2, which is related to the increased Nb content and decreased Al content promoting O phase precipitation within the B2 matrix, and restrains the formation of martensite and gas-pores in the weld.
3. The laser-welded joint properties can be significantly improved by using Nb filler metal. The strength (724 MPa) and elongation (5.1%) of the joint increased by 119.4% and 168.4% compared with those of the joint without filler metal. The joint strength reaches 81.7% of BM strength (886 MPa). It is favorable to use pure Nb filler metal for improving the mechanical properties of laser-welded Ti₃Al-Nb alloy joints.

Author Contributions: L.W. designed and conducted the study and wrote the manuscript. D.S. is the leader of the research group and helped to review the manuscript. H.L. defined the tasks and scope of the project in which the present work is framed, and she also provided technical guidance during all the stages of development of the present work and helped to prepare the presentation of the results and to explain them. X.G. helped to understand the experimental results. C.S. helped the author to carry out the different experimental tests.

Funding: This research received financial support from the National Natural Science Foundation of China (51805205).

Acknowledgments: The authors would like to thank the National Natural Science Foundation of China for financial support. They would also like to thank Zhang Yan from Jilin University for his assistance in this experiment.

Conflicts of Interest: The authors declare no conflict of interest.

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