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Relationships between microhardness, microstructure, and grain orientation in laser-welded joints with different welding speeds for Ti6Al4V titanium alloy

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Abstract: The microhardness curve trend and its relationships with microstructure and misorientation were analyzed to enhance the comprehension of the microstructure and mechanical property of micro-areas in Ti6Al4V laser-welded joints with different welding speeds. The microhardness measured on the fusion line (H_m) is the highest from the weld center to the base metal. H_m increases with increasing weld width in a welded joint and increasing degree of the non-uniformity in all studied welded joints. The microhardness decreases from the weld metal to the base metal with decreasing amount of martensite α' and increasing amount of original α phase. When the microstructure is mainly composed of martensite α' , the microhardness changes with the cooling rate, grain size of the martensite, and peak values of the fraction of misorientation angle of the martensite in a wide weld metal zone or weld center at different welding speeds, whereas the difference is small in a narrow weld metal zone.

Key words: microhardness; microstructure; misorientation angle; non-uniformity; welding speed

1 Introduction

Ti6Al4V titanium alloy is widely used in aerospace, petrochemical, biomedical, and numerous other industries owing to its low density, high specific strength, good toughness, high fatigue strength, and superior biocompatibility [1-3]. High energy beam welding and additive manufacturing, such as laser beam welding (LBW) [4-7], electron beam welding [8-10], selective laser melting (SLM) [11-12], and electron beam melting (EBM) [13], of Ti6Al4V titanium alloy have gained tremendous interest owing to its extensive use and smaller distortion compared to that obtained with conventional methods, e.g., tungsten inert gas (TIG) welding [14,15].

The microstructure, microhardness, grain orientations and residual stress were studied in Ti6Al4V high energy beam welding [16–20].

Among the different microstructures of Ti6Al4V titanium alloy under different heat treatment conditions, the yield and tensile strengths of the bimodal microstructure were the highest; the ductility of the equiaxed microstructure was the highest; and the fatigue strength of the lamellar microstructure was the highest [21]. The Vickers microhardness and tensile strength of the martensite α' in the weld metal are higher than those of the original α in the base metal [22,23]. HONG and SHIN [24] stated that the microhardness measured on the fusion line was higher than that of the weld center in Ti6Al4V laser-welded joints, whereas according to LIU et al [25] the microhardness in the weld metal remained relatively unchanged. This difference was probably caused by the differing welding parameters or plate thicknesses. Welding parameters have significant impacts on the microstructures, microhardness, mechanical properties, and welding defects of the welded

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joints [26-28]. HONG and SHIN [24] used a multi-physics model to calculate the martensite content in the heat affected zone (HAZ) and weld metal, which proved that the welding speed affected the martensite content and microhardness. Grain orientations of the different regions of a welded joint are related to the transformation process of microstructures during cooling [29-31]. LIU et al [32] observed that the cooling rate increased, grain changed from block to columnar, amount of the sub-grain boundaries in α grains increased, amount of α/α grain boundaries in a single primary β grain decreased, and the β to α transformation changed from diffusion transformation to massive transformation when the welding speed increased.

The thermal process of SLM essentially involves the rapid heating and cooling of a local region, similar to the thermal process of laser beam welding, leading to a similar microstructure in the remelted zone which consists primarily of α' martensite [33]. In comparison, the microstructure obtained with EBM includes the lamellar microstructure composed of original α and β owing to the lower cooling rate of EBM versus SLM [34,35]. BRANDL et al [36] found out that the hardness of the as-built blocks decreased from $HV_{0,1}$ 420 at the top to $HV_{0,1}$ 320 at the bottom because of the more homogeneous material at the bottom than the top. PRASHANTH et al [37] joined SLM components by friction welding to form a large SLM structure. The microhardness of the weld metal increased as the size of the martensite α' platelets decreased [37]. The microhardness of Ti6Al4V SLM samples is usually higher than that of Ti6Al4V laser-welded samples, which results from the low heat input of SLM, whereas the same low heat input will cause the lack of penetration in laser-welded joint.

In the present work, the distributions of the microhardness, microstructure, and grain orientation of a welded joint from the weld metal to the base metal were studied. The influences of the welding speed and microhardness gradient on the microhardness were examined. The relationships of the microhardness, microstructure, and misorientation were also investigated.

2 Experimental

Rolled and annealed Ti6Al4V titanium alloy

plates with dimensions of 120 mm × 50 mm × 4 mm were butt-welded by autogenous laser welding. Before welding, the plates were soaked in 10 wt.% NaOH aqueous solution for 10 min and in 5 vol.% hydrofluoric acid and 30 vol.% nitric acid aqueous solution for 2 min. The laser welding equipment includes an IPG YLS–10000–S2T laser and a YASKAWA robot. The shielding gas was argon with purity of 99.99%. The welding parameters are listed in Table 1.

Table 1	Welding	parameters	during	laser	welding
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Welding	Focal	Welding	Shielding gas
power/	position/	speed/	flow rate/
W	mm	$(\text{m} \cdot \text{min}^{-1})$	$(L \cdot min^{-1})$
4500	0	0.8 1.6 2.4	40

The metallographic specimen with dimensions of 14 mm \times 10 mm \times 4 mm and the electron backscattered diffraction (EBSD) specimen with dimensions of 7 mm \times 10 mm \times 4 mm were located 20 mm away from the arcing point to avoid the unstable arc striking region. The metallographic and EBSD specimens were prepared using standard metallographic techniques and electrolytic polishing. The electrolyte for electrolytic polishing contained 5 vol.% glacial acetic acid and 95 vol.% ethanol. For the metallographic specimen, the voltage was 31 V and electrolytic time was 120 s. For the EBSD specimen, the voltage was 28 V and electrolytic time was 90 s. The metallographic specimen was etched for 30 s in a solution containing 2 vol.% hydrofluoric acid and 2 vol.% nitric acid and observed using a ZEISS Axio Imager M2m optical microscope. The EBSD specimen was analyzed after electrolytic polishing using a ZEISS MERLIN Compact SEM equipped with an Oxford Nordlys MAX EBSD system. The EBSD tests were operated at an accelerating voltage of 15 kV, a working distance of about 20 mm, and a step size of $0.3 \mu m$.

The Vickers microhardness test was performed on the prepared metallographic specimen using a Vickers THVS-50 digital microhardness tester. The test load was 500 g, the load time was 15 s, and the spacing between two microhardness points was 0.2 mm. The microhardness distribution of the weld center (Vertical in Fig. 1) from the top surface to the bottom surface was tested. Moreover, the microhardness distributions were tested at different depths (0.5, 1.5, 2.5, and 3.5 mm from the bottom



Fig. 1 Observed and tested cross-section of welded joint with welding speed of 1.6 m/min

surface in Fig. 1) from the weld center line (WCL) to the base metal. In Fig. 1, the irregular yellow lines represent grain boundaries of the prior β phase, black lines represent the locations of microhardness test, red dashed line represents the fusion line, red dotted line represents the fully transformed HAZ (FTHAZ)/partially transformed HAZ (PTHAZ) boundary, and red solid line represents the HAZ/base metal boundary.

3 Results

3.1 Microstructures of Ti6Al4V laser-welded joint

As can be seen from Fig. 2, the microstructure of the base metal (Fig. 2(a)) is composed of a white

equiaxed original α phase and black intergranular original β phase. The microstructure of the weld metal (Fig. 2(b)) is mainly martensite α' . The martensite α' is acicular and separated by the grain boundaries of the prior β phase. During cooling of the weld metal, parallel primary martensites nucleate and grow until they reach the grain boundaries. Moreover, short secondary martensites grow in a direction perpendicular to that of the primary martensites. The secondary martensites stop growing when they encounter grain boundaries or primary martensites. HAZ can be divided into FTHAZ (Fig. 2(c)) and PTHAZ (Fig. 2(d)) based on the microstructure. The highest temperature of the FTHAZ, which is near the weld metal, exceeds the β transus temperature. The microstructure of the FTHAZ mainly includes an acicular martensite α' and a blocky transformed α . The transformed α in HAZ has clear irregular grain boundaries, and it typically occurs in the vicinity of the martensite α' . The highest temperature of the PTHAZ, which is near the base metal, is lower than the β transus temperature. Besides the acicular martensite α' and blocky transformed α , original α and β phases are present in the PTHAZ. From the weld metal to the base metal, the amount of martensite α' decreases, the amounts of the original α and β phases increase, and the amount of transformed α initially increases and then decreases.



Fig. 2 Microstructures of Ti6Al4V laser-welded joints: (a) Base metal; (b) Weld metal; (c) FTHAZ; (d) PTHAZ

3.2 Grain orientation of Ti6Al4V laser-welded joint

Figure 3 shows the grain orientation and grain boundary maps of the grain misorientations above

 10° of various micro-areas in the Ti6Al4V laserwelded joint. In the base metal, the green β phase is distributed at the grain boundaries of the red original α phase. There is a little β phase in the



Fig. 3 Grain orientation (a, b, c, d) and grain boundary (a', b', c', d') maps of Ti6Al4V laser-welded joint: (a, a') Base metal; (b, b') Weld metal; (c, c') FTHAZ; (d, d') PTHAZ

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PTHAZ. The red α grains with a large aspect ratio in the weld metal and HAZ are martensite. In a prior β grain, the martensites in the weld metal are nearly perpendicular to each other. The length of the martensite in the PTHAZ is shorter than that in the FTHAZ. In the weld metal, the areas with sparse grain boundaries are grain boundaries of the prior β . The microstructure of the grain boundaries of the prior β phase is α phase, and the less grain boundaries than martensites indicate that the α phase at the grain boundaries of prior β is transformed to α instead of martensite α' .

The distributions of the grain misorientations above 10° of the α phase in the base metal and weld metal are significantly different, as shown in Fig. 4. The misorientations of the original α in the base metal are in the range of 10°–94°. The peak of the fraction of grain misorientations of the original α in the base metal is around 10°, and the peak value is lower than 0.05. The weld metal is mainly composed of martensite, and the misorientations of the α phase in the weld metal are in the ranges of 9.5°–13.5°, 52.5°–66.5°, and 86.5°–91.5°. The peaks of the fraction of grain misorientations in the

weld metal are around 11°, 60°, 64°, and 90°, and the peak values at 60° and 64° are much higher than those of the fraction of the grain misorientations of the original α . Martensite follows the Burgers orientation relationship (BOR) of $\{0001\}_{\alpha}//\{110\}_{\beta}$, $\langle 1120 \rangle_{\alpha} / / \langle 111 \rangle_{\beta}$ with prior β grains during the rapid cooling of the laser welding [38,39], which makes the misorientations of martensite distribute in a specific range. The distributions of the grain misorientations of the HAZ are similar to those of the weld metal because the HAZ contains a large amount of martensites. The differences between the fraction of the grain misorientations of the HAZ and the weld metal are shown in the peak values at 60° and 90° of the HAZ being lower and higher than those of the weld metal, respectively. The misorientations of the PTHAZ are not only in the ranges of 9.5°-13.5°, 52.5°-66.5°, and 86.5°-91.5°, as can be seen in the inset of Fig. 4(d), which indicates that the original α exists in the PTHAZ. From the weld metal to the base metal, the peak value of the fraction at 10° of the misorientations increases, the peak value at 60° decreases, and the peak values at 64° and 90° initially increase and then decrease.



Fig. 4 Grain misorientations of α phase in different micro-areas of welded joint: (a) Base metal; (b) Weld metal; (c) FTHAZ; (d) PTHAZ

3.3 Non-uniform microhardness distribution of welded joint at three different welding speeds

Figure 5 shows the microhardness distributions at different depths (0.5, 1.5, 2.5, and 3.5 mm from the bottom surface) of the welded joint obtained at a welding speed of 1.6 m/min. The scatter points in Fig. 5 are the measured values of the microhardness, and the curves in Fig. 5 are the fitted curves using a linear fit in the weld metal and the DoseResp function in the HAZ and base metal. There is little difference in the microhardness of the weld center in a welded joint at different depths. The microhardness curve of the upper weld metal (3.5 mm from the bottom surface) rises from the WCL to the fusion line, and those of the lower (0.5 mm from the bottom surface) and middle weld metals (1.5 and 2.5 mm from the bottom surface) are flat. Moreover, the rise in the microhardness curve causes the microhardness near the fusion line of the upper weld metal to be higher than those of the lower and middle weld metals having a flat microhardness curve. Figure 5(b) shows the



Fig. 5 Microhardness distributions at different depths with welding speed of 1.6 m/min (a) and characteristic microhardness curves (b)

characteristic microhardness curves of the Ti6Al4V laser-welded joint. The microhardness of the martensite α' is higher than that of the original α because more lattice imperfections are caused by the fast cooling rate in martensite, which results in a higher microhardness for the weld metal than the base metal. In Fig. 5(b), the flat solid line in the weld metal represents the microhardness curve of the lower and middle weld metals in Fig. 5(a), whereas the rising dashed line in the weld metal represents the microhardness curve of the upper weld metal in Fig. 5(a). $H_{\rm m}$ is the maximum microhardness in the microhardness curve, which is located near the fusion line. In the PTHAZ, the microhardness decreases gradually from the FTHAZ/PTHAZ boundary to the base metal.

Figure 6 shows the relationships between the weld width and slope of the microhardness curve of the weld metal at various depths of the welded joints at three different welding speeds. When the weld width is smaller than 1.1 mm, the slope of the microhardness curve of the weld metal is zero, which implies that the microhardness curve of the weld metal is flat, and $H_{\rm m}$ is as high as the microhardness of the weld center. When the weld width is larger than 1.1 mm, the slope of the microhardness curve of the weld metal rises with increasing weld width in general. The slope of the microhardness curve of the weld metal is positive, which implies that the microhardness increases from the WCL to the fusion line. The points that are marked in Fig. 6 will be explained subsequently in the discussion section.

Figure 7 shows non-uniform distributions of the microhardness and microhardness gradient of the Ti6Al4V laser-welded joints at three different



Fig. 6 Relationships between weld width and slope of microhardness curve of weld metal



Fig. 7 Contours of welded joints: (a) Microhardness; (b) Microhardness gradient

welding speeds. The microhardness near the fusion line of the upper joint is the highest in a welded joint. The highest microhardness increases with the increasing welding speed. When the welding speed is 0.8 m/min, the microhardness of the welded joint is uniform in thickness. When the welding speeds are 1.6 and 2.4 m/min, the microhardness of the upper HAZ is higher than that of the lower HAZ. The degree of the non-uniformity of the microhardness in thickness increases with the increasing welding speed. The microhardness gradient is the absolute value of the changing rate of the microhardness, which represents the degree of non-uniformity in the transverse direction of the welded joints. From the WCL to the base metal, the microhardness gradient of the PTHAZ is the highest. As the welding speed increases, the microhardness gradient increases, which implies that the degree of non-uniformity in the transverse direction increases. When the welding speed is high, e.g., 1.6 and 2.4 m/min, the microhardness gradient of the upper HAZ is higher than that of the lower HAZ.

3.4 Relationship between microhardness, microstructure, and grain misorientation

3.4.1 Comparison of different micro-areas in weld metal

The microstructure in the weld metal is mainly martensites α' , but the microhardness of different micro-areas in the weld metal is significantly

different when the weld width is large. Thus, four micro-areas in the weld metal, which were 1.0, 1.5, 2.0 and 2.6 mm from the WCL and near the top surface of the welded joint with a welding speed of 0.8 m/min, were selected to analyze the microstructure morphology and grain misorientations when the weld width is wide. As the distance from the WCL increases, the amount of martensite boundaries increases, and the martensites become finer after ignoring the prior β grain boundaries (Fig. 8(a)). The misorientation angles in Fig. 8(b) are basically in the range of those of martensites because the weld metal is mainly composed of martensites. As the distance from the WCL increases, the peak value of the fraction at 60° of the misorientation decreases by 0.066, whereas that at 64° increases by 0.076 and 90° increases by 0.031. Figure 8(c) shows that the microhardness increases and the grain area of the martensite decreases exponentially from the weld center to the fusion line in the wide weld metal zone.

Three micro-areas in the weld metal, which were 0.1, 0.5 and 0.8 mm from the WCL and near the bottom surface of the welded joint at a welding speed of 1.6 m/min, were selected to analyze the microstructure morphology and grain misorientations when the weld width is narrow. As the distance from the WCL increases, the amount of martensite boundaries has no significant change (Fig. 9(a)). As can be seen from Fig. 9(b), the peak value of the fraction at 60° of the misorientations is



Fig. 8 Misorientations and microhardness of micro-areas with different distances from WCL in wide weld metal zone: (a) Grain orientation and grain boundary maps; (b) Grain misorientation; (c) Microhardness and grain area

lower than 0.20, and it has no tendency when the distance from the WCL increases. The peak values of the fraction at 64° and 90° decrease with the increasing distance from the WCL, which is opposite with that in the wide weld metal zone. Figure 9(c) shows that there is little change in the microhardness and the grain area of the martensites from the weld center to the fusion line in the narrow weld metal zone.

3.4.2 Comparison of weld center at different welding speeds

As the welding speed increases from 0.8 to 2.4 m/min, martensite laths are finer (Fig. 10(a)), the peak value of the fraction at 60° of the misorientations of the martensites decreases, whereas the peak values at 64° and 90° increase (Fig. 10(b)), the grain area of the martensites decreases from 31 to 19 μ m² and the microhardness increases from HV 359 to 378 (Fig. 10(c)). The increase in the welding speed leads to an increase in

the cooling rate [40], which shortens the growth time of the martensites, makes the martensites finer, and changes the peak values of the grain misorientations of the martensites.

4 Discussion

4.1 Microstructures and phase transformations in Ti6Al4V laser-welded joint

Ti6Al4V alloy is a two-phase titanium alloy including the original α and β phases. The crystal structure of the original α phase is hexagonal close packed (HCP), and its lattice parameters are a=b=0.292 nm and c=0.466 nm [38]. Al is rich in the α phase, which is stable at low temperatures such as 20 °C. The crystal structure of the original β phase is body centered cubic (BCC), and its lattice parameters are a=b=c=0.320 nm [41,42]. V is rich in the β phase, which is stable at high temperatures such as 1000 °C. During the solidification of the



Fig. 9 Misorientations and microhardness of micro-areas with different distances from WCL in narrow weld metal zone: (a) Grain orientation and grain boundary maps; (b) Grain misorientation; (c) Microhardness and grain area



Fig. 10 Misorientations and microhardness of weld center of welded joints at different welding speeds: (a) Grain orientation and grain boundary maps; (b) Grain misorientation; (c) Microhardness and grain area

liquid metal, the liquid phase transforms into the prior β phase, and then the prior β phase transforms into the α variants. The type of the α variant that forms during cooling depends on the cooling rate.

When the cooling rate is higher than 410 °C/s, the prior β phase transforms into fully martensite α' via a diffusionless transformation during cooling. When the cooling rate is in the range of 20–410 °C/s, the

prior β phase transforms into the martensite α' phase via a diffusionless transformation and into the transformed α via a massive transformation during cooling. When the cooling rate is lower than 20 °C/s, the prior β phase transforms into the equilibrium α phase by a diffusion transformation during cooling. The transformed α nucleates successively at the prior β boundaries, martensite plates adjacent to the prior β boundaries, and individual martensite α' plates within the prior β grains with decreasing cooling rate [43].

In the weld metal, the metal transforms into the liquid phase during heating. And the liquid transforms into the prior β phase, then transforms into martensite α' and transformed α after rapid cooling during laser welding. In the FTHAZ, the metal transforms into the prior β phase during heating. And the prior β phase converts into martensite α' and transformed α phase during rapid cooling. In the PTHAZ, some original α transforms into the prior β phase during heating, then the prior β phase turns into martensite α' and transformed α phase during rapid cooling, and some original α and β phase do not transform into the prior β phase during heating because the temperature is lower than the β transus temperature.

4.2 Relationships between microhardness, weld width, and non-uniformity

Figure 11(a) exhibits the relationship between the microhardness (H_m) and weld width. In a welded joint, H_m increases with increasing weld width. In the welded joints at different welding speeds, the highest and lowest $H_{\rm m}$ both increase with increasing welding speed. The microhardness point which is marked with a circle in Fig. 6 is the microhardness of the upper weld metal at a welding speed of 2.4 m/min where a large microhardness gradient exists. The microhardness point which is marked with a box in Fig. 6 is the microhardness of the middle weld metal at a welding speed of 0.8 m/min where a small microhardness gradient exists, which indicates that the microhardness of the weld metal is not only relevant with the weld width, with the microhardness gradient. but also Figure 11(b) exhibits the relationship between the microhardness and non-uniformity. H_m increases with the increasing microhardness gradient and welding speed, which implies that H_m increases with the increasing degree of non-uniformity.



Fig. 11 Relationships between weld width, microhardness gradient and H_m : (a) Weld width and H_m ; (b) Microhardness gradient and H_m

4.3 Relationships between microhardness, microstructure, and misorientation

There are three types of misorientation angles in the welded joints of Ti6Al4V: 2-8.029° of substructure boundaries (Type 1); 8.029-13.029°, 57.5-65.762° and 87.5–92.5° of α/α grain boundaries in one parent β grain (Type 2); 13.029-57.5°, 65.762-87.5° and 92.5-94° of α/α grain boundaries in two parent β grains (Type 3) [32]. Subgrains and misorientation angles less than 10° are not considered in this study. The grain boundaries of the martensites are α/α grain boundaries in one parent β grain. And the misorientation angles of the weld metal and FTHAZ, which are composed of mainly martensite, basically fall in the range of the misorientation angles of Type 2. The grain boundaries of the original α phase are α/α grain boundaries in two parent β grains. Thus, the fraction of Type 2 could represent the amount of the martensite, and the fraction of the other misorientation angles except Type 2 could represent the amount of the original α phase. Figure 12 shows the microhardness, fraction

of type 2 and other misorientation angles in the weld metal, FTHAZ, PTHAZ, and base metal. The differences in the microhardness and fraction of misorientations in the weld metal and FTHAZ are small. From the FTHAZ to the base metal, the microhardness decreases by about HV 40 with decreasing amount of the fraction of type 2 misorientation angles and increasing amount of the other, which means that the microhardness decreases with decreasing amount of the mattersite α' and increasing amount of the original α .



Fig. 12 Fraction of misorientation angles and microhardness of different micro-areas

When the microstructure is mainly martensite, such as in the weld metal, the microhardness still fluctuates. The microhardness near the fusion line is higher than that of the weld center when the weld width is wide. This is because the differences in the cooling rate, grain size, and the fraction of misorientation of the weld center and fusion line are much larger for a wide weld metal zone than a narrow weld metal zone. Moreover. the microhardness of the weld center increases with increasing welding speed. The cooling rate of the fusion line is higher than that of the weld center [24], and the cooling rate increases with increasing welding speed [40]. When the microhardness increases obviously in the weld metal, the cooling rate increases, the amount of grain boundaries of the martensites increases, the grain area of the martensites decreases; the peak value of the fraction at 60° of the misorientations of the martensite decreases, whereas those at 64° and 90° increase. Figure 13 shows the relationships of the microhardness and peak values of the fraction of misorientations of the martensite in the weld metal zone. As the microhardness increases, the peak value of the fraction at 10° of misorientations

is nearly invariable, where as that decreases at 60° , increases at 64° and slightly increases at 90° .



Fig. 13 Relationship between fraction of misorientation and microhardness in wide weld metal zone

5 Conclusions

(1) From the weld metal to the base metal, the microhardness decreases with decreasing amount of the martensite α' phase and increasing amount of the original α phase, and the amount of the transformed α phase first increases and then decreases. Further, the peak value of the fraction at 10° of the misorientation angles increases, the peak value at 60° decreases, and the peak values at 64° and 90° first increase and then decrease from the weld metal to the base metal.

(2) The microhardness measured on the fusion line, H_m , is the highest from the weld center to the base metal. As the welding speed increases, the highest microhardness of the welded joint and the degree of non-uniformity of the microhardness in thickness and transverse direction both increase. H_m increases with increasing weld width in a welded joint and increasing degree of non-uniformity in different welded joints.

(3) When the weld width is small, the change is small between the microhardness, grain area of the martensite and peak value of the fraction of the misorientations in the weld metal. When the weld width is wide, the microhardness increases, the grain area of the martensites decreases, and the peak value of the fraction at 60° of the misorientations of the martensite decreases, whereas the peak values at 64° and 90° increase with increasing cooling rate from the weld center to the fusion line or when welding speed increases.

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不同焊速下 Ti6Al4V 激光焊接头 显微硬度、组织和晶粒取向联系

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摘 要:为了提高对 Ti6Al4V 钛合金激光焊接接头中各微区的显微组织和微区力学性能的理解,对不同焊接速度 下接头中显微硬度曲线的分布趋势及其与显微组织和取向差之间的关系进行研究。从焊缝中心到母材,熔合线附 近的显微硬度值(*H*_m)最高。在同一接头中,*H*_m随着焊缝宽度增加而增加;在不同接头中,*H*_m随着不均匀度增加 而增加。从焊缝到母材,随着马氏体含量减少和初始α相含量增多,显微硬度逐渐降低。当显微组织主要由马氏 体组成时,在较宽的焊缝中或在不同焊速接头的焊缝中心,显微硬度随着冷却速率、马氏体大小和马氏体的晶界 取向差而变化;焊缝较窄时,冷却速率等引起的显微硬度的差异较小。