Metallurgical aspects of joining commercially pure titanium to Ti-6Al-4V alloy in a T-joint configuration by laser beam welding

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Abstract

The present paper focuses on the metallurgical and microstructural characterization of the laser beam-welded T-joints between commercially pure titanium (CP-Ti) and Ti-6Al-4V alloy. The weld regions were comprehensively studied and the mechanisms leading to the final morphology within each weld region were described. The link between microstructural features and local mechanical properties was demonstrated. Owing to different constitution, the responses of the two titanium alloys to thermal cycles imposed by laser welding are completely different. A strong interface with no dilution zone between the two alloys was observed. The cooling rate during the welding process is high enough for diffusionless martensitic transformation in the Ti-6Al-4V part of the fusion zone. In contrast, no evidence of martensite was found in the CP-Ti because of low solute content and, consequently, much higher critical cooling rate. Plausible reason for some controversy found in the literature on the resulting transformation products after laser processing of CP-Ti was given. The present findings might have important industrial implications because careful microstructural characterization revealed the real position of the skin fusion line, which is of great importance for fulfillment of the weld quality criteria.

Keywords Laser beam welding · Titanium alloys · T-joint · Microstructure · EBSD

1 Introduction

The reduction of the environmental impact of aviation has nowadays become one of the major concerns for aircraft manufacturers. New and highly efficient approaches and innovations in airframes, engines, and aerodynamics are required for emission reductions in order to compensate for the rapid growth in air travel. From the airframe perspective, the improvement of the

overall aircraft performance primarily depends on the introduction of advanced materials and novel manufacturing processes. In this context, the concept of producing the hybrid structures combining dissimilar lightweight materials enhances the product design flexibility and cost-effectiveness of the manufacturing process. The introduction of dissimilar joints allows the different materials to be utilized in an efficient and functional manner based on their specific properties. It is important to maximize the contribution of each material of the joint to ensure the overall optimal performance in terms of mechanical properties.

The application of laser beam welding (LBW) has been comprehensively studied and considerably developed over the last three decades. This process was established as a suitable method for the welding of dissimilar metals with minimum distortion [1]. LBW uses a high-intensity laser beam as the heat source and can be used over a wide range of powers and velocities to produce welds covering a range of depths with sufficient penetration. The major advantages of the LBW technique are high welding speeds (up to 10 m/min), narrow fusion zones (FZ) and heat-affected zones (HAZ), low heat input, and high flexibility of the process from the practical point of view. The keyhole mechanism of laser welding is capable of producing welds with high depth-to-width ratio

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compared to those made by electron beam welding (EBW). It is important to provide metallurgical compatibility of the welded materials in the dissimilar joints made by the LBW.

In the present study, the LBW of dissimilar joints between commercially pure titanium (CP-Ti, ASTM Grade 2) and Ti-6Al-4V alloy (ASTM Grade 5) in a T-joint configuration was investigated. A new and creative approach to combine the high formability of the CP-Ti skin with the high strength of Ti-6Al-4V stringer allows the production of a structure exhibiting excellent mechanical properties and complying with the stringent regulations of the aircraft industry [2]. Owing to this tailor-engineered approach, different parts of the same weldment possess their own mechanical and physical properties required at specific locations. A major drawback of titanium alloys is the high cost of the raw material and the fabricated part. From this perspective, the LBW technique is a near-net-shape process that enables an increase in the fly-to-buy ratio and the overall cost-effectiveness.

With a guaranteed minimum yield strength of 275 MPa and good ductility and formability, CP-Ti is the workhorse of α titanium alloys for industrial applications. Its yield strength is comparable to those of annealed austenitic stainless steels [3]. This alloy has sufficiently small total alloying additions, such that it almost does not respond to heat treatment, but this characteristic contributes to the excellent weldability of this alloy. Ti-6Al-4V alloy belongs to the group of dual-phase ($\alpha + \beta$) titanium alloys. At room temperature, it contains around 10–15% of the β phase, depending on the thermo-mechanical processing route. Ti-6Al-4V is the most widely used titanium alloy. The aerospace industry accounts for more than 80% of its usage [3].

Titanium and its alloys can be welded by various types of fusion welding including tungsten inert gas (TIG) welding [4], plasma arc welding (PAW) [5], EBW [6], and LBW [7–11]. Nowadays, TIG welding is the most widespread technology for joining titanium alloys; however, low welding speeds, high heat input, and distortion can result in the inferior weld quality compared to EBW and LBW [4, 12]. Owing to numerous advantages, laser welding is a perspective alternative to conventional welding techniques. Mechanical properties of the welded joints are always related to the microstructure in the weld zone, which is primarily affected by the transformation kinetics after solidification. Several researchers have investigated the microstructure-property relationship in the laser beam-welded and electron beam-welded CP-Ti [6, 9–11]. Nevertheless, to date, there has been little agreement on the resulting transformation products after fusion welding of CP-Ti. Lathabai et al. investigated the microstructure and mechanical properties of CP-Ti joints produced by the keyhole TIG welding [4]. They observed a coarsening of the equiaxed β grains in the FZ of the weld. Within prior β grains, internal substructure consisting of parallel α plates was detected. It was shown that $\beta \rightarrow \alpha$ transformation has occurred by a nucleation and growth process, resulting in the formation of a

Widmanstätten microstructure. Wu et al. studied the evolution of the microstructure and microtexture during the EBW of Grade 1 alloy [6]. The authors have concluded that besides the grain coarsening, crystal orientation is an important factor which can affect the mechanical properties of the FZ. The increase in yield strength observed for the welds was attributed to the texture strengthening due to the suppression of prismatic slip in the weld zone. It has been shown by Li et al. that the purity of the shielding gas can affect the resulting microstructure of the laser beam-welded seam of CP-Ti [9]. The authors have shown that additions of oxygen in the shielding gas reduce the coarse-grained serrated α phase and increase the fine-grained acicular α in the weld metal. Elmer et al. conducted the in situ mapping of the phases that exist in the HAZ during the arc welding process of CP-Ti [12]. They come to the conclusion that the $\beta \rightarrow \alpha$ transformation appears to take place by a massive transformation mechanism. In the work of Liu et al., the mechanical properties and strengthening mechanisms in laser beam welds of pure titanium were investigated [11]. It has been found that the strengthening mechanism in the FZ is ascribed to the substructure strengthening and the solute solution strengthening. In contrast to pure titanium, it is generally agreed that martensitic $\beta \rightarrow \alpha$ transformation occurs in the FZ of the laser-welded Ti-6Al-4V alloy upon solidification [7, 8].

In spite of numerous studies on the microstructural characterization of laser-welded CP-Ti and Ti-6Al-4V, the effect of laser processing on the dissimilar CP-Ti/Ti-6Al-4V joint has not been reported yet. The aim of this study is to characterize the welding seam from the metallurgical perspective and to gain insights into the mechanisms of the seam formation. Combining two different titanium alloys with different transformation kinetics in one weld makes the welding metallurgy significantly more complicated. Mechanical properties of the joints produced by fusion welding are strongly correlated to the resulting microstructure of the weld zone after solidification. An in-depth understanding of the mechanisms and kinetics of the microstructural transformations within the fusion zone is required for the anticipation of the weld quality and strength. This paper presents the detailed analysis of the weld zones and links the resulting microstructure within the weld seam with the welding characteristics. The distribution of local mechanical properties, as represented by microhardness mapping, showed the significant effect of microstructural transformations on the resulting quality and strength of the joint.

2 Materials and experimental procedure

Two different titanium alloys have been used for the LBW of dissimilar T-joints in the current work. The material of the skin was commercially pure titanium (CP-Ti, ASTM Grade 2,

AMS 4902E [13]). The material of the stringer was Ti-6Al-4V alloy (ASTM Grade 5, AMS 4911F [14]). All the materials were delivered in the form of $500 \times 500 \text{ mm}^2$ sheets with a thickness of 0.8 mm. Prior to welding, the sheets were cut into the coupons of required geometry. The nominal chemical composition of the alloys from the corresponding standards is shown in Tables 1 and 2. The alloying-element contents determined by energy-dispersive X-ray (EDX) analysis are also provided for Ti-6Al-4V alloy. A comparison of the basic mechanical properties of the two titanium alloys is given in Table 3. The values correspond to the rolling direction of the sheet. In spite of having almost equal elastic modulus, tensile strength of Grade 5 is higher by more than a factor of two. At the same time, the ductility of the Ti-6Al-4V alloy is considerably lower than that of the CP-Ti.

The welding equipment consisted of an 8-kW continuous-wave ytterbium fiber laser YLS-8000-S2-Y12 (IPG Photonics Corporation) integrated with a six-axis KUKA KR30HA industrial robot. A collimation lens with 120-mm focal length, a focusing lens with 300-mm focal length, and a process fiber with a diameter of 200 μm were employed to produce a focal spot diameter of approximately 500 μm . The center wavelength of the fiber laser was 1070 nm. The divergence half-angle of the focused multimode beam was 29.5 mrad, and the resulting beam parameter product $\text{BPP} = 7.4 \text{ mm mrad}$. Prior to welding, the faying edges of the specimens were machined, ground, and thoroughly cleaned with ethanol to remove any surface oxides and contaminants. A detailed description of the welding process, the equipment used, shielding conditions, and optimization of the process parameters has already been reported in our previous study [2]. The parameters finally chosen and employed for the welding of coupons are listed in Table 4. Local shielding by Ar flow was provided by the nozzles mounted on the optical head. The welding direction was parallel to the rolling direction (RD) for both the skin and the stringer. The schematic illustration of the LBW process is shown in Fig. 1. The focal spot with a diameter of 0.5 mm was located with a 0.3-mm offset relative to the skin surface.

Transverse cross-sections were cut from the stable middle region of the joint for metallographic examination and microhardness testing. After sectioning, the samples were mounted, ground, and polished using an oxide polishing suspension (OP-S). The wet grinding was done in a few steps using normal SiC-grinding papers with up to 4000 grit size. Significantly different mechanical properties of the alloys, combined in one dissimilar joint, make the preparation of the sample for metallography quite challenging. The relatively high ductility of the CP-Ti is the main reason for much worse surface quality of the skin with numerous scratches after grinding. To achieve a sufficient surface quality of the CP-Ti part of the sample, a special metallographic preparation is required. This problem can be overcome by the use of modified OP-S with the addition of hydrogen peroxide and ammonia.

Microstructural observations were performed using both inverted optical microscopy (OM) Leica DMI 5000 M and scanning electron microscopy (SEM) JEOL JSM-6490LV. Prior to light microscopy, the specimens were etched to unveil the microstructural features. Two types of reagents were used for etching—Kroll's etchant (2% HF, 6% HNO_3 , 91% distilled water) and Weck's color etchant (100 mL of water, 25 mL of ethanol (96%), and 2 g of ammonium bifluoride NH_4HF_2) [15]. The most common chemical etchant for titanium alloys is Kroll's reagent; however, its effect was more pronounced on Grade 5 than on CP-Ti. The clearest microstructure image of CP-Ti was obtained by the color tint etch using the Weck's reagent and polarized light microscopy (PLM). For SEM investigations, a mirror-like OP-S polished surface was used. The SEM microstructure observations and microtexture analysis of the joints were conducted using electron backscatter diffraction (EBSD). The EBSD measurements were performed at an acceleration voltage of 30 kV, an emission current of 75 μA , a working distance of 12–14 mm, a step size of 0.5 μm , and a sample tilt angle of 70°. For the orientation calculation, the generalized spherical harmonic series expansion (GSHE) method was applied on the basis of triclinic sample symmetry. EDX analysis was used for the local chemical composition determination. For EDX analysis, SEM was operated at an acceleration voltage of 15 kV, a working distance of 10 mm, and a live time of 150 s. The quantitative analysis was based on the standard ZAF method of correction.

Microhardness profiling was conducted in order to characterize the local mechanical properties in the weld zone. Transverse cross-sections of the samples for microhardness testing were prepared in the same manner as discussed for the microstructural evaluations. The Vickers microindentation hardness test was carried out using a Zwick/ZHU0.2/Z2.5 universal hardness testing machine and testXpert software. The samples were tested with a 200 g load applied for 15 s according to ASTM E384-11. The indentation spacing was kept at 150 μm in the CP-Ti and 100 μm in the Ti-6Al-4V alloy, to provide the minimum recommended distance between test point.

3 Results and discussion

3.1 Macroscopic appearance of the welded joints

The visual inspection of the obtained laser weldments showed bright silver metallic surfaces, indicating that stable Ar-shielding was provided from all the sides of the joint and along the whole length. All the joints welded with the optimal parameter set were fully penetrated. Sufficient penetration of the skin and stringer could be recognized even through visual inspection. The presence of melted and re-solidified material

Table 1 Chemical composition (wt.%) of the Ti-6Al-4V alloy (Grade 5)

	Al	V	Fe	Si	N	C	H	O	Ti
AMS 4911F [14]	5.5–6.75	3.5–4.5	< 0.3	–	< 0.05	< 0.08	< 0.015	< 0.20	Bal.
Measured	6.18	3.51	0.21	0.48	–	–	–	–	88.18

from the back side of the stringer can be regarded as a simple indicator of the fully penetrated joint. It is important to note that this indicator is only valid when relatively thin sheets are welded. Kashaev et al. [8] reported that for 2.5-mm-thick sheets, the joint could have not sufficient penetration in spite of the presence of re-solidified material from both sides of the stringer.

Owing to 0.3-mm offset of the laser spot relatively to the skin, the laser power is almost completely absorbed by the stringer surface. Nevertheless, sufficient penetration was provided owing to the high intensity of the laser beam. High energy density in the focal spot leads to the formation of a channel filled with vaporized material, known as a keyhole [16]. In the present work, two effects can be considered as results of the presence of the keyhole during the LBW—first, the shape of the fusion zone is considerably elongated in the direction of the laser beam, and second, excessive porosity was usually observed in the region of the keyhole tip (Fig. 2). Apparently, the underlying reason for porosity formation is keyhole instability and the resulting entrapment of some portion of the shielding gas within the FZ.

A typical transverse cross-section of the joint welded with optimal process parameters is shown in Fig. 2. Several weld zones can be visibly distinguished by the variations in local grain size and shape. These variations are the inevitable result of the microstructural transformations taking place during the laser welding process. Obviously, two kinds of base material—designated as the skin BM (CP-Ti) and stringer BM (Ti-6Al-4V)—were identified as the regions where no microstructural changes took place. These zones begin at a remote distance from the weld zone, where the temperature is not sufficient for any microstructural transformations of titanium. The fusion zone of the joint, where—by definition—the material was melted and subsequently solidified, can be subdivided into stringer FZ and skin FZ. The interface between these two zones is clearly visible, even in the low-magnification image shown in Fig. 2. The stringer FZ consisting of Ti-6Al-4V alloy was distinguished by the fine needle-like morphology. The skin FZ located under the stringer FZ was characterized by coarse grains of irregular shape. The stringer HAZ

adjacent to the FZ represents the transition between the needle-like morphology in the FZ and the stringer BM.

It is important to mention the rationale behind the consideration of the coarse-grained region of the skin as the “skin FZ” instead of “skin HAZ.” Indeed, it is not obvious from the general appearance (Fig. 2) that this region was melted and then re-solidified. First of all, no grain coarsening should be expected in the HAZ of laser weldments. Typically, high heating and cooling rates prevent diffusion-controlled grain growth. Thus, coarse grains in this region could only arise upon the solidification of the molten pool. Additional evidence for the fact that the investigated region was melted was found in the more detailed observations of the skin FZ. The layer immediately adjacent to the skin BM consists of coarse planar grains that grew epitaxially from the crystalline structure of the BM. All primary grains that solidify at the base-metal/weld-metal interface exhibit such epitaxial growth. This is a unique mechanism peculiar only to the nucleating of the first grains of solid metal from the molten pool of the weld. Epitaxial solidification soon gives way to competitive growth and resulting coarse globular microstructure in the center of the FZ. The epitaxially grown initial layer is clearly visible in Fig. 3a, obtained by the color tint Weck’s etchant and cross-polarized light.

The absence of pronounced HAZ in the skin can be also explained by taking into account the features of the alloy used for the skin material. CP-Ti, being an α alloy, has much less total alloying additions than Ti-6Al-4V; consequently, it essentially does not respond to heat treatment [17]. Hence, thermal cycles—even with high peak temperatures—do not alter the original microstructure, and the transition region between the skin FZ and the skin BM is not formed.

It is important to note that the Grade 2/Grade 5 interface should not be confused with the depth of fusion into the skin. In this context, Fig. 2 can be somewhat misleading because the clear boundary between the two materials can be regarded as the fusion line. In fact, the interface between the skin FZ

Table 2 Nominal chemical composition (wt.%) of the CP-Ti (Grade 2) [13]

	C	Fe	H	N	O	Other	Ti
AMS 4902E	< 0.08	< 0.3	< 0.015	< 0.03	< 0.2	< 0.3	99.2

Table 3 Static mechanical properties of the materials used (average values with standard deviation from three tested specimens)

	CP-Ti	Ti-6Al-4V
Modulus of elasticity, E, GPa	99.8 ± 4.1	99.5 ± 2.5
Yield strength, $R_{p0.2}$, MPa	371.8 ± 5.7	972.5 ± 5.8
Ultimate tensile strength, R_m , MPa	433.8 ± 4.2	1045.0 ± 5.9
Elongation at break, A, %	31.2 ± 0.6	15.4 ± 2.1

Table 4 Optimal parameter set for the LBW of dissimilar CP-Ti/Ti-6Al-4V T-joints

Parameter	Value
Laser power, W	650
Welding speed, m/min	3.5
Incidence position a , mm	0.3
Incidence angle, α	25°
Focus position, mm	0

and stringer FZ is just the border between two titanium alloys with different chemical composition. The variations in alloying-element content lead to significantly different final microstructures after solidification. This strong microstructural gradient from the fine needle-like morphology in the stringer FZ to the coarse-grained microstructure in the skin FZ is shown with higher magnification in Fig. 3b. As discussed, the fusion line in the skin is much deeper than the Grade 2/Grade 5 interface. Moreover, Fig. 3a shows that the depth of fusion is almost equal to the skin thickness. This is the inevitable result of the application of fusion-welding technology to the joining of such thin sheets. The depth of fusion into the skin is one of the most important acceptance criteria for T-joints obtained by any type of fusion welding. Without a careful microstructural characterization, the interface line between the two titanium alloys could be mistakenly regarded as insufficient penetration in the skin and lead to rejection of the applied optimal process parameter set.

3.2 Microstructural analysis

Five distinct regions of the welding seam, as shown in Fig. 2, were carefully studied by OM and SEM. OM is a fast and simple way for microstructural characterization; however, it has limited magnification (up to $\times 1000$ for the microscope used in the current research) and the results obtained are usually more qualitative in nature. Thus, the general appearance and the weld zones were primarily investigated by OM and the regions of interest were then observed with high

Fig. 1 Schematic illustration of the LBW process **Fig. 2** Transverse cross-section of the welded T-joint with a typical location of porosity (OM image, Kroll's etchant)

magnification using SEM. EBSD measurements with special OIM software enabled the measurement of the average grain size and the quantitative texture analysis. In the following section, the microstructure of each weld region is described in detail.

3.2.1 Titanium Grade 2 base material

The microstructure of the as-received CP-Ti sheet consists of equiaxed α grains with a small volume fraction of the β phase (typically less than 3%), as shown in Fig. 4a. A non-uniform distribution of the average grain size was observed in the investigated CP-Ti sheets. The low-magnification image shown in Fig. 4b reveals the variation of the grain size along the thickness of the sheet. It is clear that the grains are considerably smaller in the middle section of the sheet than near the surface. This non-homogeneous microstructure is probably a result of the non-uniform distribution of plastic deformation during the rolling process.

For quantitative grain mapping, EBSD technique was used. The color-coded orientation map of the region in the middle zone of the sheet is shown in Fig. 5a. The color in the crystal orientation map is based on a color-coded inverse pole figure, in which different colors represent different crystallographic orientations. The average grain size found by the OIM software was $7.9 \pm 2.8 \mu\text{m}$.

The type and intensity of texture is mainly attributable to the working practice—i.e., deformation degree and deformation temperature during the rolling process. Texture strengthening is one of the few available options for the strengthening of single phase α titanium alloys. As shown in Fig. 5b, the texture of the CP-Ti sheet in the as-received condition has the maximum concentration of basal poles lying along the direction that connects the sheet normal and the transverse direction. The point of maximum basal pole concentration is

Fig. 3 **a** Transverse cross-section of the laser-welded T-joint (OM image, cross-polarized light, Weck's tint etchant). **b** CP-Ti/Ti-6Al-4V interface (OM image, cross-polarized light, Weck's tint etchant)

inclined by about 30° away from the sheet normal toward the transverse direction (TD). It is typical pole figure for CP-Ti that was unidirectionally rolled [9, 17].

3.2.2 Titanium Grade 5 base material

The as-received BM microstructure of the 0.8-mm-thick Ti-6Al-4V sheet consists of globular α grains with the average grain size of $3.2 \pm 1.1 \mu\text{m}$ and an intergranular retained β , as shown in Fig. 6a. The volume fraction of β phase was estimated to be around 10%, which is typical for Ti-6Al-4V alloy. Figure 6a shows the OM image of the microstructure in the middle section of the sheet. The bright regions are the equiaxed α grains, and the dark regions are the intergranular β grains distributed at α grain boundaries. The most influential microstructural parameter on the mechanical properties of the fully equiaxed microstructure is the α grain size [17].

The crystallographic texture of α phase—which primarily develops through the deformation process—can influence the mechanical properties of the final product [18]. The microtexture analysis, as represented by orientation map and

pole figures in the cross-section plane, is shown in Fig. 6b, c. Note that TD has different direction compared to that shown in Fig. 5 because the Grade 5 sheet is perpendicular to the Grade 2 sheet in a T-joint. It should be kept in mind that the normal to the cross-section in Fig. 6a–c coincides with the RD of the sheet. The regions of analysis shown in Fig. 6a–c are taken from the middle section of the sheet. The laminated morphology of the material in TD direction, which is clearly visible in Fig. 6b, apparently originates from the plastic deformation of the lamellar structure during the rolling process. The (0002) pole figure shows that basal planes are aligned in both the RD (center of the pole figure) and the normal direction of the sheet (thickness direction). According to the investigation by Salem [18], the alignment of some portion of the basal planes in the RD leads us to conclude that the material could be cross-rolled.

3.2.3 Skin FZ

High peak temperatures, severe temperature gradients, and rapid thermal fluctuations occur as the laser beam passes

Fig. 4 Microstructure of CP-Ti. **a** OM image of the middle section of the sheet, Kroll's reagent. **b** Grain size distribution along the thickness (Weck's reagent, polarized light)

Fig. 5 Texture analysis of the CP-Ti. **a** Crystal orientation map. **b** (0002) and (11–20) pole figures in the cross-section plane

through a material. The boundary of the fusion zone—the so-called fusion line—corresponds to the position of the solidus temperature ($T_s = 1604^\circ\text{C}$ for titanium). During the welding process, the material of the FZ is melted and subsequently resolidified. The microstructural transformations taking place upon solidification of the molten pool usually lead to the resulting morphology that is quite different from the original one prior to the melting. As a result, the position of the fusion line can be clearly distinguished—in both the skin and the stringer (Fig. 2).

The skin FZ microstructure shown in Fig. 7a, b was characterized by coarser grains than those in the skin BM (Fig. 5a). As discussed in Section 3.1, in the close proximity of the fusion line, the grains are elongated due to preferential grain growth in the direction of the heat flow (epitaxial layer). In the

center region of the FZ, globular grains of irregular shape dominate the microstructure. There is no evidence of any internal substructure within these coarse grains from the OM observations. However, EBSD analysis reveals that the large grains consist of a number of small subgrains, which also have a slightly elongated shape (Fig. 7a, b). The misorientation in the crystal direction between the subgrains within one large grain is negligibly small; consequently, they are marked by the same color on the EBSD map. Nevertheless, the internal structure of the coarse grains is clearly visible in Fig. 7b, where the regions with low misorientation angle within one large grain are shown by slightly different shades of one color. The misorientation angle between the internal subgrains was lower than 5° . A basic understanding of the welding metallurgy and the main mechanisms of transformation kinetics is needed

Fig. 6 Equiaxed microstructure of the as-received Ti-6Al-4V sheet. **a** OM image. **b** EBSD orientation map in the cross-section. **c** (0002) and (11–20) pole figures in the cross-section plane

Fig. 7 Microstructure of the skin FZ. **a** Overview EBSD orientation map of the skin FZ. **b** Middle region of the skin FZ with higher magnification

to explain this morphology. Compared to the rolled BM, the texture in the skin FZ was much weaker and close to random. Therefore, the pole figures are omitted.

The fusion welds solidify from the edge of the fusion zone and the solid/liquid interface moves in toward the weld centerline. Once the solidification is complete, the local temperature of the material starts to gradually decrease. Solid-state transformations occurring during the cooling period alter the local microstructure and depend on the alloy composition. Upon cooling, titanium undergoes an allotropic change from a high-temperature body-centered cubic phase (β) to a low-temperature hexagonal close-packed structure (α) [17]. Thus, the $\beta \rightarrow \alpha$ transformation occurs on the trailing side of the weld zone as the single-phase β region transforms to α . The temperature of equilibrium $\beta \rightarrow \alpha$ allotropic transformation depends on the alloying elements and is equal to $882 \pm 2^\circ\text{C}$ for pure titanium [17]. In Grade 2 titanium alloy, the $\beta \rightarrow \alpha$ transition temperature is altered by the presence of iron, oxygen, and other impurities [3]. Moreover, the temperature range for the transformation depends on the cooling rate and gradually moves toward a lower temperature with increasing cooling rate. It was reported that CP-Ti could be treated as a titanium alloy with an $\alpha/(\alpha + \beta)$ equilibrium temperature of $885 \pm 5^\circ\text{C}$ and a $\beta/(\beta + \alpha)$ equilibrium transus temperature of $915 \pm 5^\circ\text{C}$ [12].

The transformation of the β phase to the α phase in Ti alloys can be categorized by three distinctive transformation modes, depending on the cooling rate and the alloy composition. At low cooling rates, diffusion-controlled nucleation and growth process takes place. At cooling rates that are faster than a critical cooling rate, the martensitic reaction occurs. The martensite transformation is diffusionless in nature and involves the cooperative movement of atoms through a shear

type process [19]. At intermediate cooling rates, massive transformation can occur by nucleation and short-range diffusion jumps across the massive/matrix interface [20, 21]. The martensitic and massive transformations are both initiated and completed without any change in the composition between the parent and the product phase. Very often, a careful consideration of the constituent morphology and crystallography is needed to differentiate martensite from massive products.

Fully martensitic transformation in pure titanium takes place above a critical cooling rate of about 3000 K/s [3, 22]. However, it is reported that the critical cooling rate of martensitic reaction for CP-Ti can largely differ from the values for pure titanium [12, 20]. The difference arises mainly due to the presence of iron, which is a strong β -stabilizer. A comprehensive study on the mechanisms of continuous cooling $\beta \rightarrow \alpha$ transformation behavior of CP-Ti was performed by Kim and Park [20], who report that extremely fine lathlike features of martensite dominate the microstructure at cooling rates faster than about 600 K/s. This significant drop in the critical cooling rate from 3000 to 600 K/s was attributed to the addition of 0.07 wt.% of iron. Thus, the transformation kinetics strongly depends on the alloy constitution; even negligibly small additions of β -stabilizing elements can alter the final transformation products. Comparing the coarse-grained microstructure of the skin FZ obtained in the current work with the fine needle-like morphology found by Kim and Park [20], we can conclude that—in the present study—the cooling rates during the LBW process are not sufficient for diffusionless martensitic transformation. Any evidence of fine lamellar or acicular morphology was found by neither OM nor SEM. Apparently, the iron content of the alloy investigated is much lower than that reported in [20]; as a result, the critical cooling rate is close to the value for pure titanium.

Fig. 8 Comparison of the skin FZ microstructure with that after water quenching. **a** OM image of the skin FZ. **b** Morphology of the water-quenched CP-Ti

There is still some controversy over the resulting transformation products after the laser processing of CP-Ti. It is generally established that laser treatment considerably modifies the shape and size of the grains. However, no clear evidence of martensite is observed in the laser beam-welded butt joints of CP-Ti by Li et al. [9]. Similar results are reported by Amaya-Vazquez et al. [23] for CP-Ti treated by laser surface remelting. In contrast to the aforementioned studies, a few researchers observe clear acicular martensitic microstructure in the laser surface-treated CP-Ti [24, 25]. The authors believe that these contradictory results are related to the different content of β -stabilizing elements and, consequently, different transformation kinetics of the investigated alloys.

Thus, the coarse grains in the skin FZ shown in Fig. 7a, b are prior β grains formed upon solidification. During the $\beta \rightarrow \alpha$ transformation, the cooling rates are not sufficiently high to provide martensitic diffusionless $\beta \rightarrow \alpha'$ transformation. Within these large prior β grains, slightly elongated α grains are formed with approximately the same crystal orientation. Additional TEM measurements should verify whether the products of massive transformation are present in the structure.

In order to compare the microstructure of the skin FZ with a real martensite of the CP-Ti alloy used in this work, water

quenching of as-received Grade 2 sheets was conducted. The material was heated up to 1200 °C and then water-quenched. The resulting microstructure is shown in Fig. 8b. As we can see, it is characterized by large irregular regions (size about 50–100 μm) containing packets of small, almost parallel α plates (thickness about 0.5–1.0 μm). This is a typical martensitic morphology of pure titanium and very dilute titanium alloys [17]. As we can see, there is a great difference between the martensitic structure and the morphology obtained in the skin FZ (see Fig. 8a). Thus, we can conclude that the cooling rate in the skin FZ is considerably lower than that upon water quenching. As a result, α grains are formed by diffusion-controlled reaction rather than by martensitic mechanism.

3.2.4 Stringer FZ and HAZ

In spite of similar thermal cycles experienced in the stringer FZ, its microstructure is completely different to that of the skin FZ. The underlying reason for this is the different constitution and, consequently, transformation kinetics of the Ti-6Al-4V alloy. Similar to CP-Ti, there are three basic mechanisms of $\beta \rightarrow \alpha$ phase transformations—diffusional transformation [26], displacive/martensitic transformation [19], and massive

Fig. 9 Microstructure of the stringer FZ. **a** OM image (Kroll's etchant). **b** EBSD orientation map

transformation [27]. Depending on the alloy composition and the cooling rates from the β -phase field, the phase transformations in Ti-6Al-4V can take place in these three pathways. Ahmed and Rack investigated the effect of different cooling rates on microstructural transformations in the Ti-6Al-4V alloy [28]. They reported that cooling rates above 410 K/s were required to achieve fully martensitic microstructure and a massive transformation was observed between 410 and 20 K/s. Massive mechanism was gradually replaced by diffusion-controlled Widmanstätten α formation at cooling rates lower than 20 K/s. In their investigation, Ahmed and Rack found that at cooling rates lower than 410 K/s, grain boundary α morphology was formed upon cooling.

The martensitic structure of the Ti-6Al-4V alloy differs from the typical water-quenched martensite of CP-Ti shown in Fig. 8b. In titanium alloys with high solute content, the products of diffusionless martensitic transformation have acicular morphology. The acicular martensite consists of an intimate mixture of individual α' plates, each having a different variant of the Burgers relationship. Generally, martensitic plates contain a high dislocation density and sometimes twins [17]. The microstructure of the stringer FZ is shown in Fig. 9a, b. Within large prior β grains, the FZ predominantly consists of a needle-like acicular α' martensitic structure. Since no evidence of secondary α formation at prior β grain boundaries was observed (Fig. 9b), we can conclude according to the results of Ahmed and Rack that the cooling rates in the stringer FZ are high enough for $\beta \rightarrow \alpha'$ diffusionless transformation of the Ti-6Al-4V alloy. In contrast to CP-Ti, the relatively high solute content of Grade 5 reduces the critical cooling rate and the final microstructure is fully martensitic. Similar to the skin FZ, no pronounced texture components were found in the pole figures.

In contrast to the non-heat-treatable CP-Ti, Ti-6Al-4V alloy has a wide range of microstructures achievable through an appropriate type of heat treatment. The content of the β phase

strongly depends on the temperature, and annealing even at temperatures lower than β -transus (around 1000 °C at low cooling rates [17]) can lead to considerable changes in phase composition and microstructure. This makes the Ti-6Al-4V alloy more sensitive to thermal fluctuations in the proximity of the welding zone. As a result, a clear HAZ is formed in the region adjacent to the fusion line in the stringer (Fig. 2). Stringer HAZ displays the transition region between the acicular morphology in the FZ and the globular structure in the BM.

HAZ is usually divided into two subregions based on the β -transus temperature. In the HAZ adjacent to the FZ (near-HAZ), the temperatures exceed the β -transus during the LBW. Consequently, this region consists mostly of the transformed acicular morphology, with a small amount of embedded prior α grains. Because the temperatures in the HAZ adjacent to the BM (far-HAZ) are lower than the β -transus, its microstructure is very similar to that of the BM. Thus, the fraction of the transformed martensitic structure gradually decreases from 100% in the HAZ near the fusion line to zero in the BM. Comprehensive analysis of the HAZ formation in the laser beam-welded Ti-6Al-4V alloy can be found elsewhere [29].

3.3 Local chemical composition

The alloying-element contents (wt.%) of the base materials are listed in Tables 1 and 2. Taking into account the detection limits of the EDX analysis, values lower than approximately 0.5% should be taken as qualitative. For this reason, no measured data are shown for CP-Ti. Interestingly, some amount of silicon (Si) is found in the Ti-6Al-4V alloy in spite of the absence of this alloying element in the standard. Apparently, this addition was deliberately made in order to increase the creep resistance of Ti-6Al-4V. It is widely recognized that a moderate addition of Si can significantly improve the creep behavior of Ti alloys [30].

Fig. 10 **a** Location of the EDX measurement line. **b** Profile of Al and Ti content in the stringer FZ/skin FZ transition region

From the metallurgical perspective, the fusion welding of dissimilar materials is a complicated process, primarily due to the formation of a transition zone between the metals. The intermetallic compounds formed in the transition zone can have a strong influence on crack sensitivity, ductility, and susceptibility to corrosion. The dissimilar joints can be made successfully only if there is a mutual solubility of the two alloys. In the present work, the welded materials are two titanium alloys; consequently, they have equal melting temperatures and thermal expansion coefficients. The Al and V contents of the Grade 5 alloy are too low for the formation of any intermetallic phases.

A considerable difference in chemical composition leads to the strong spatial gradients of alloying-element content at the Grade 2/Grade 5 interface. In order to study the distribution of alloying additions in the transition region, EDX line measurements were performed, as shown in Fig. 10a. The measurement line was perpendicular to the interface and has the total length of 600 μm with 0.3- μm steps. Figure 10b shows the profile of relative proportions of Al and Ti in the transition region. For quantitative analysis, the data from Tables 1 and 2 can be used. The main conclusion that can be drawn here is that the dilution zone between the two alloys is negligibly narrow (around 10 μm). The Al content drops very steeply during the transition from Grade 5 to Grade 2 alloy. This result is consistent with the high cooling rates encountered during the LBW process. The characteristic time of a few tenths of a second for which the material is subjected to high temperatures is not sufficient for diffusion processes. As a result, no clearly pronounced dilution zone was formed; the transition from one alloy to another was characterized by steep gradients of chemical composition.

3.4 Microhardness distribution

The 2D contour plot of the microhardness distribution over the welding zone is shown in Fig. 11a. The average microhardness of the skin BM (CP-Ti) is $156 \pm 6 \text{ HV}_{0.2}$. Ti-6Al-4V BM has a significantly higher average microhardness of $314 \pm 7 \text{ HV}_{0.2}$, which is consistent with the results of the tensile tests (Table 3). As shown in Fig. 11a, the stringer FZ exhibits the highest average microhardness of approximately $395 \pm 10 \text{ HV}_{0.2}$ and it decreases abruptly as the distance from the fusion line increases. The steep increase in microhardness in the stringer FZ is correlated with the local changes in microstructure activated during the LBW process. The occurrence of maximum hardness in the FZ is attributed to the formation of a strong martensitic structure resulting from high cooling rates upon solidification. The acicular α' phase produced by the diffusionless transformation exhibits higher strength and lower ductility, which are related to the fine size of the martensitic plates and high defect density [17].

In contrast to the acicular morphology of the stringer FZ, no considerable changes of microhardness were observed in the skin FZ. This result is consistent with the microstructural observations, because no evidence of martensitic morphology was found. Nevertheless, the fact that microhardness does not drop in the skin FZ leads us to conclude that there is a fine substructure within the large prior β grains. Otherwise, one should expect the decrease in hardness by the Hall-Petch mechanism because of grain coarsening after solidification. Similar substructure strengthening mechanism was reported by Liu et al. [11]. As shown in Fig. 11a, strengthening effect of the final products during cooling is significantly more pronounced in the stringer FZ compared to the skin FZ. In this

Fig. 11 Microhardness ($\text{HV}_{0.2}$) distribution over the welding zone. **a** 2D contour plot of microhardness distribution. **b** The profile of averaged microhardness along the y-axis

study, it is hard to quantify the effect of texture strengthening separately from the other strengthening mechanisms like Hall-Petch, substructure or formation of the martensitic α' phase. Microhardness values reflect the combined effect from all the mechanisms together. Comprehensive study on how the microtexture components can affect the local slip behavior in the welding zone can be found elsewhere [10, 18].

Figure 11b shows the profile of the average microhardness along the Y-axis of the T-joint (Fig. 11a). The boundary between the skin and the stringer is shown by the dashed line. A slight increase in microhardness can be seen in the skin FZ. It is followed by the steep spatial gradient of microhardness at the skin/stringer interface. The stringer FZ exhibits maximum hardness values of up to 400 HV_{0.2}. After this, the curve decreases to the average microhardness value of the stringer BM. The increased deviation of the microhardness values in the skin FZ can be related to the grain coarsened morphology with random crystal orientation in this region. Since the indentation size is smaller than the typical grain size, the resulting HV values differ from grain to grain depending on its orientation.

Although the five distinct regions of the joint exhibit completely different microstructural characteristics, they are combined in one weld and, consequently, their properties and effect on the weld seam are interconnected. The link between different weld zones is reflected by the microhardness distribution, which shows the full integrated picture. Thus, microhardness map represents the combined effect of all the transformations that took place locally upon solidification.

Since both the tensile strength and hardness are attributed to slip within grains, they are thought to have the strongest correlation with each other. The empirical relationship between yield stress, ultimate tensile strength, and hardness HB or HV is of great importance for material characterization. It was shown that the linear relationship between UTS and HV found for steels also holds for Ti alloys [31]. Thus, hardness measurements—being significantly easier from a practical point of view—can be considered as an indicator of the local strength of the material. In this regard, the distribution shown in Fig. 11a is beneficial from the design perspective because the stringer, playing the role of stiffener in a structure, has higher strength and load-carrying capability.

4 Conclusions

Fully penetrated dissimilar Ti-6Al-4V/CP-Ti T-joints were successfully produced by the LBW process. The present research mainly focuses on the characterization of the laser beam-welded seams from the metallurgical perspective and links the microstructural features with the distribution of local mechanical properties. The following conclusions can be made:

1. Five distinct weld zones were distinguished by variations in local microstructure—namely the stringer BM (Grade 5), stringer HAZ, stringer FZ, skin FZ, and skin BM (Grade 2). A strong interface between the acicular morphology of the stringer FZ and the coarse prior β grained structure of the skin FZ was observed. By the presence of epitaxially grown layer, it was shown that the fusion line in the skin is located considerably deeper than the CP-Ti/Ti-6Al-4V transition. This finding is very important from the perspective of fulfillment of the acceptance criteria.
2. Skin BM (CP-Ti) was characterized by equiaxed microstructure, however, with non-uniform distribution of the grain size along the thickness. Stringer BM has typical globular microstructure after rolling process and recrystallization. Stringer FZ has an acicular α' martensitic morphology because the cooling rates are higher than the critical cooling rate for this alloy. Skin FZ consists of large irregular prior β grains with no internal features observed in OM. However, SEM observations reveal an internal structure consisting of small slightly elongated α grains. No evidence of martensite was found in the skin FZ, apparently due to the low content of β -stabilizing elements and the high critical cooling rate.
3. The distribution of the main alloying elements in the welding zone was obtained by EDX spectroscopy. The transition between the two alloys was characterized by steep spatial gradients of chemical composition. No clearly pronounced dilution zone was observed at the skin/stringer interface.
4. Microhardness mapping gave some insight into the distribution of the local mechanical properties over the welding zone. CP-Ti and Ti-6Al-4V have average microhardness values of 156 ± 6 HV_{0.2} and 314 ± 7 HV_{0.2} respectively. The highest value of microhardness of up to 400 HV_{0.2} was observed in the stringer FZ as a result of hardening due to martensitic transformation. Skin FZ exhibits an average microhardness that is approximately equal to that of the as-received CP-Ti BM. Steep spatial gradients of hardness are consistent with the distribution of alloying-element content at the CP-Ti/Ti-6Al-4V interface.

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