Improvement weldability of dissimilar joints (Ti6Al4V/Al6013) for aerospace industry by laser beam welding

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Abstract
The discovery of new metal alloys and the technological advancement in welding processes are key resources for the aerospace industry to obtain cost reduction and better reliability. Thus, welded joints of dissimilar materials such as aluminum and titanium alloys have been explored due to their combined low density and high mechanical performance. Otherwise, welding of dissimilar metals may present deleterious factors to the welded joint as the formation of intermetallic and/or brittle second phase and residual stress. This project investigates the weldability of dissimilar welded joint (Al6013/Ti6Al4V) by laser beam welding. The approach will be done in terms of mechanical properties and microstructural characterization. For this purpose, optimal laser offset from the joint line and the related heat input has been found. It was observed that offset controls the amount of the intermetallic compound layer in the fusion zone. Large pores were observed on the Al side of the weld metal when the offset is zero. The microstructure on the aluminum side consisted of $\alpha$-Al grains and the dispersed precipitates. Heat input and offset are also influenced in the volumetric fraction of the precipitates. Martensite $\alpha$ and secondary acicular $\alpha$ phase were found in the titanium side. Furthermore, intermetallic compound of TiAl base phase such as TiAl, Ti$_3$Al$_4$, and Ti$_2$Al$_3$ was formed. Tensile strength of welded joint was 60% of the Al alloy. In addition, for the same offset and higher heat input, there was an increase in the hardness of the interface.

Keywords Intermetallic compounds · Dissimilar butt joint · Fiber ytterbium · Laser welding · Titanium alloy · Aerospace industry

1 Introduction
The use of dissimilar Al/Ti joints to obtain structures and components is a promising choice for the aerospace industry. In addition, there is also a growing application of dissimilar joints in the medical, shipbuilding, consumer goods, electronics, and automotive sectors [1–3]. The combinations between the properties of aluminum and titanium alloys have advantages such as increased corrosion resistance, better mechanical properties, and reduced weight and cost of the structure or component [4, 5]. In this context, there are widely publicized studies on the replacement of riveting on aircraft fuselages by welding panels reinforced with stringers (welding of skin-stringer joints) to reduce the weight of the structure. Specifically, titanium-aluminum alloys are successfully applied to honeycomb sandwich and seat-track of aircraft panels [6].

Likewise, a study was made on the feasibility of hybrid structures consisting of aluminum, titanium, and carbon fiber reinforced with polymeric matrix (CFRP) used in automobiles and aircraft [7], for example, in the protection panels of the lower wing skin panels and cargo door of the Boeing C-17, as well as in the junction of the wing constituted by Ti alloy with part of the fuselage of Al alloy [8]. Recently, research has been done on the use of Al/Ti joints in the frames of aero bodies and missiles [9].

The welding or brazing of dissimilar materials, in particular Al/Ti, is a challenging process, considering the different properties between the alloys, such as crystalline structure, melting temperature, thermal conductivity, and thermal expansion...
Another important feature is that both Ti and Al alloys have high chemical activity, the result of which is that the oxide formed is not conducive to the junction interface between the two alloys (Al/Ti) and may produce slag in the inside the weld [11] and form TiO_2 and Al_2O_3 on the surface. It also results in layers of intermetallic compounds (IMCs) and/or other deleterious phases at the welded joint interface as a consequence of the chemical, physical, and metallurgical differences of the dissimilar metals and according to the Al/Ti phase diagram, these being the main BMIs: T3Al, TiAl, TiAl_2, and TiAl_3 [2, 12–14].

Several welding/brazing processes and techniques have been studied in order to control the precipitation of IMCs in the joint region. Vacuum diffusion joining process was used in Al/Ti alloys applying aluminization to the Ti plate surface, resulting in increased formation time for the intermetallic layer at the Al/Ti. By controlling the diffusion parameters, it was possible to reduce the formation of intermetallic [15]. Furthermore, Havlík et al. conducted a study on the weldability of the Ti6Al4V/AA 6061 dissimilar joint obtained by the electron beam process by varying the beam offset. The author reported the presence of Ti3Al, TiAl, Ti5Al11, and Ti3Al5 as well as volume fraction of IMCs controlled by the offset variation [16].

Solid-state welding processes, including explosion welding, friction stir welding [17–20], friction welding [21], and ultrasonic welding [22], were also evaluated for the Al/Ti scenario; however, these process techniques have limitations as to joint configuration and structure size. All the welding or soldering-brazing processes studied still need improvement due to the complexity of the factors involved because they may have low joint strength, dimensional limitation, high cost, and impracticability for large-scale production.

Laser welding has been widely applied due to the high energy density, fast processing speed, less thermal deformation, and low distortions of the parts. In addition, proper heat input and laser offset are capable of suppressing metallurgical reactions, such as formation of IMCs [1, 23, 24]. However, the weld bead’s appearance and the stability of the process depend on several factors such as the low wettability/spreading of aluminum on titanium.

Casalino et al. evaluated the formation of IMCs in Al/Ti dissimilar joints achieved by fiber laser offset welding where no filler metal was used and also identified the influence of the morphology of the IMCs with the mechanical properties of the joint [25, 26].

Song et al. evaluated the mechanical and microstructural properties of Ti6Al-4V/A6061 joint by laser brazing without filler metal. Good surface quality was obtained, and through the process parameters, it was possible to reduce the intermetallic layer formation, followed by an increase in the tensile strength by 64% compared to base metal [27]. Vaidya et al. performed laser welding, without filler metal, between AA6056 and Ti6Al4V alloys using a U-shaped chamfer technique on one interface. By using optimal welding parameters, it was possible to reduce the intermetallic layer, TiAl3, and reduce crack propagation in fatigue conditions [28].

This study proposes to investigate and optimize the mechanical performance of the dissimilar butt joint obtained by the laser fiber welding, between the AA 6013-T4 aluminum alloy and the Ti6Al4V titanium alloy, both used in the aerospace industry. The influence of welding energy and offset (distance between laser spot and plates interface) on the microstructure, intermetallic compounds, and mechanical properties of the welded joint was identified.

### 2 Materials and methods

#### 2.1 Materials

The experiments were carried out using two different metal alloy plates, the aluminum class AA 6013-T4 and the titanium alloy Ti6Al4V - Grade 5, both plates with dimensions of 100 mm × 50 mm × 1.6 mm. The T4 description of the aluminum plate corresponds to a tempering designation, in which the aluminum alloy was subjected to the solubilization cycle, rapid cooling, and natural aging. The titanium sheet was used as received (obtained by the rolling process and subjected to the annealing heat treatment according to ASTM B265). The rolling direction was preserved for both the aluminum alloy plate and the titanium alloy plate. Table 1 shows the chemical composition and Table 2 shows the mechanical properties of the alloys used in the research.

#### 2.2 Welding procedure

The welding experiments were carried out using the laser welding equipment, IPG, YLR-2000 model located at the Multi-User Laboratory for the Development of Laser and Optics Applications of the Photonics Division (EFO) of the

| Table 1 Chemical composition of base metal, wt.% |
|-----------------------------------------------|
| Al | C  | Fe  | Ti  | V   | Mn  | Si  | Cu  | Mg  | Cr  | Zn |
|----|----|-----|-----|-----|-----|-----|-----|-----|-----|----|
| Ti6Al4V* | 5.72 | 0.06 | 0.37 | Base | 4.00 | -   | -   | -   | -   | -  |
| AA 6013** | Base | 0.50 | 0.10 | -   | 0.80 | 1.00 | 1.00 | 1.20 | 0.10 | 0.25 |

*According to the manufacturer
**Ref.[29]
The equipment offers 2-kW continuous laser of medium power, the active medium being a fiber coated with ytterbium, output with 50-μm diameter, beam parameter product (BPP) equal to 6.3 mm.mrad, and 5 m in length. The laser wavelength is 1.07 μm, beam quality (M2) of approximately 1, Gaussian distribution intensity TEM00 [33], and a beam diameter of 0.1 mm. The details of the processing station where the laser is inserted are shown schematically in Fig. 1.

During welding process, argon was used as a shielding gas with a flow rate of 18 l/min (as illustrated in Fig. 2). The welding parameters are presented in Table 3.

### 2.3 Microstructure characterization

Samples for microstructural characterization were taken from cross-section perpendicular to the welding direction. All of them were mounted in bakelite. Hence, mounted samples were wet ground on progressively finer grades of SiC (240, 400, 600, and 1200, respectively) impregnated emery paper using copious amounts of water both as a lubricant and as a coolant. Afterward the samples were polished using 3- and 1-μm diamond solution, respectively. After a short time, the polished samples were etched using Kroll reagent (10 ml of hydrofluoric acid, 45 ml of nitric acid, and 45 ml of water) for times varying from 1 to 5 s. The etched sample’s surface was analyzed with an optical microscope, Zeiss Axio Imager Z2m, and photographed. For scanning electron microscopy (SEM), an electron microscope SEM-FEG Inspect50 with secondary, backscattered electrical detectors, EDS X-ray spectrometer and backscattered electron diffraction EDAX camera (EBSD-TEAM), of the Laboratório de microscopia Eletrônica e de Força Atômica (LabMicro – USP), was used.

### 2.4 Hardness test

The hardness analysis of the welded joint was obtained by microindentation carried out by Vickers hardness test (HV 0.2) using an automated Dura-Scan 50 (Emcotest, Kuchl, Austria) to map the hardness distribution in a pre-selected area.

### 2.5 Tensile test

Tensile specimens were cross-sectioned perpendicular to the welding direction and the tensile tests were accomplished on small size specimens according to ASTM E 8M - 04 standard [34]. For this purpose, it was used a testing machining EMIC DL-10000 with a maximum load capacity of 100 kN.

Five samples were tested for each welding condition, of which just 3 was used for average calculation to evaluate the tensile strength of the joint. Some specimens were discarded due to unforeseen events during the tensile test (e.g., slippage of the claw). The terminology “CDPXX” was used to identify each sample for the tensile test, where CDP means “specimen” and XX is the numbering which corresponds to the numerical order in which the specimen was subjected to the tensile test, for example, CDP01 (see Fig. 14).

### Table 2 Mechanical property

| Alloys       | Tensile strength (MPa) | Yield stress (MPa) | Modulus of elasticity (GPa) | Elongation (%) |
|--------------|------------------------|--------------------|-----------------------------|----------------|
| Ti6Al4V [31, 32] | 895                    | 830                | 114                         | 10             |
| AA 6013 [30]   | 350 ± 2                | 230 ± 2            | 53 ± 16                     | 28 ± 1         |

![Fig. 1 Schematic diagram of processing station with fiber laser](image1)

![Fig. 2 Schematic drawing of the laser welding processing](image2)
3 Results

3.1 Weld bead profile

The joint’s cross-section images corresponding to conditions T1, T2, T6, T7, T8, and T9 are shown in Fig. 3. Only these conditions showed sufficient strength in the welded joint to not break when handled and withstand the efforts submitted during the inlay procedure. It is observed that the molten zone is more extensive on the top, and narrower on the root. This indicates greater absorption of the incident radiation on the top of the joint (side of the titanium plate). From the above, it was observed that the welding conditions adopted produced welds with full penetration for the side of the titanium alloy and met class A of the AWS D17.1 standard [35], except for the conditions T2 and T8, which weld beads showed discontinuity along the interface, with a weak joint formed on the top of the joint, as well pores in the molten area on the side of the titanium alloy.

Furthermore, the pores formed on the lower half of the weld bead (as in the case of sample T2) were caused by a non-constant geometry of the keyhole cavity [36]. This region, close to the keyhole bottom wall (RKW—rear keyhole wall) is maintained by the convection and reflection of thermal energy through the transported steam metal.

A notable behavior in the molten zone of the aluminum alloy for sample T8 was the formation of pores, especially in the center of the Al/Ti interface. As already mentioned, the fusion of the aluminum alloy took place through the thermal energy emanated by conduction and radiation from the keyhole formed in titanium. It is known that the trapping of hydrogen or a shielding gas is the main factor for the formation of pores in the weld which is due to the fast cooling speeds of the molten metal [37]. Furthermore, the aluminum alloy elements have a low boiling point and lower viscosity of the molten metal than the titanium alloy. This promotes rapid fluctuations in the liquid metal which may cause discontinuities in the Al/Ti interface of the welded joint, such as the voids and bubbles that were observed (see T2 and T8 in Fig. 3).

T9 sample, shown in Fig. 3, exhibited the most extended FZ and HAZ compared to other conditions. Also, melting of both aluminum and titanium at the Al/Ti interface of the welded joint was observed and, consequently, a better wettability of aluminum in titanium obtained. Even though condition T9 suffered the same welding energy and the same offset as condition T2, the value of the welding speed was half of T2. The welding speed and beam intensity, or laser power, affected the weld bead geometry and quality.

3.2 Effect of heat input and offset on microstructure

As mentioned above, HAZ extent, volume of the molten metal, the volumetric fraction of the phases present, and the intermetallic compounds are associated with the value of the offset and the welding energy used in each condition of the experiments. Likewise, the appearance of the weld bead is closely related to the action that the molten metal has to wet the top of the titanium alloy. This characteristic agrees with other brazing experiments on dissimilar joints [8].

Considering the cases in which the offset was assigned to the side of the titanium alloy, that is, the laser beam focused on the titanium plate, at certain distances from the Al/Ti interface, only in this phenomenon did the phenomena responsible for the formation of the keyhole occur. The aluminum was heated to a melting temperature by propagated conduction from the titanium side. Therefore, to calculate the welding energy (HI), the absorption coefficient of the titanium alloy, Ti6Al4V ($A_{bs}$ = 0.4), was used [38].

| Conditions | Power (kW) | Welding speed (mm/s) | Offset (mm) | Remarks |
|------------|------------|----------------------|-------------|---------|
| T1         | 1.2        | 33.3                 | 0.5/Ti side | Good bonding, welded at face and root only |
| T2         | 1.0        | 33.3                 | 0.5/Ti side | Weak bonding, welded at face only |
| T3         | 1.2        | 50.0                 | 0.5/Ti side | No bonding |
| T4         | 1.0        | 50.0                 | 0.5/T side  | No bonding |
| T5         | 1.2        | 33.3                 | 0.3/Ti side | Weak bonding, welded at root only |
| T6         | 1.0        | 33.3                 | 0.3/Ti side | Good bonding, welded joint |
| T7         | 1.2        | 50.0                 | 0.3/Ti side | Weak bonding, welded at face and root only |
| T8         | 1.0        | 50.0                 | 0.3/Ti side | Weak bonding, welded at root only |
| T9         | 1.0        | 16.6                 | 0.5/Ti side | Good bonding, welded joint |
| T10        | 1.5        | 16.6                 | 0.8/Al side | No bonding |
| T11        | 1.5        | 16.6                 | 0.8/Ti side | No bonding |
| T12        | 1.5        | 16.6                 | 0.5/Ti side | No bonding |
| T13        | 1.50       | 16.6                 | 0.0 (Al/Ti interface) | No bonding |
In laser welding, the welding energy ($HI$) is a function of the absorbed power ($P_A$) and the welding speed ($W_s$):

$$HI = \frac{P_A}{W_s}$$

(1)

The value of $P_A$ corresponds to the product of the intensity of the beam that hit the plate and the average value of absorptivity [39]. Table 4 presents the values of the heat input for each welding condition.

Figure 4 shows the microstructure of the welded joint that includes the cross section of the top region, the center, and the root of the weld bead. In all aluminum regions, it is possible to observe that in the region of the HAZ, near the bonding zone, there is a partially fused zone (PFZ), and there are grains with elongated epitaxial growth, as well as equiaxial grains. Columnar grains grew epitaxially from the fusion line, while equiaxial grains were formed in the liquid phase above the direction of solidification due to constitutional sub-cooling [40].

In Fig. 4a, the existence of solidification cracks along the dendritic grains in the melted area were found. Cracking by solidification occurs when residual tensile stresses (due to the action of high temperature) are induced by the internal contraction and the external displacement of the weld metal, exceeding the flow limit of the weld metal within a fragile temperature range.

The dimension of the partially fused zone was measured using imaging software (ImageJ®) on the microstructure.
images. The value of the extension of the partially molten zone, for the T1 and T2 conditions, are close (72.5 and 74 μm, respectively), which corresponds to two or three grains, according to the literature [41]. However, the average grain size in T1 is 83% larger than the average grain size in T2. This fact is associated with the different values of the welding energy in these conditions. Knowing that the increase in welding energy slows down the cooling rate and tends to increase the grain size, Fig. 4b indicates the region where there possibly is the formation of a reactive layer at the bonding interface for condition T1.

In Fig. 4c and d, we observe the image of the microstructure of the central region of the cross section to the welded joint for sample T1. Through the analysis of the images, it was found that, in the center of the weld, there was no formation of the reactive layer at the interface or any other coalescence mechanism. Likewise, this fact occurs under the conditions shown in Fig. 6. Figure 5 shows the transition region between the HAZ (coarse grains) and the molten zone of the aluminum alloy. The image shows the α-Al grains and the presence of distributed precipitates (dark spots), in addition to the growth of cellular dendritic grains.

Figure 6d highlights the discontinuities in the molten zone of the aluminum alloy, at the vicinity of Al/Ti contact interface (central region of the weld cross section, condition T8), in addition to the distinct morphology and size of the α-Al grains, when comparing them with all conditions of the same offset value. The molten zone is characterized by coarse dendritic grains, segregated interdendritic phases and with the absence (or reduction) of hardening phases [42, 43].

Figure 7a shows the base metal microstructure of the Ti6Al4V alloy. It is observed that it consists of a matrix of α phase, which corresponds to the region of light color (with compact hexagonal structure), where the β phase is dispersed, which corresponds to the dark field (with the cubic structure of a centered body) [44].

The fused zone of the titanium side was practically unchanged, being formed by coarse columnar grains of previous β. Due to the fast-cooling rate, characteristic of the laser welding process, the martensitic transformation of β occurs.

| Table 4 | Heat input values |
|---------|------------------|
| Sample | Power (kW) | Offset (mm) | Ws (mm/s) | HI (J/mm) |
| T1     | 1.20     | 0.5        | 33.3     | 14.4      |
| T2     | 1.00     | 0.5        | 33.3     | 12.0      |
| T5     | 1.20     | 0.3        | 33.3     | 14.4      |
| T6     | 1.00     | 0.3        | 33.3     | 12.0      |
| T7     | 1.20     | 0.3        | 50.0     | 9.6       |
| T8     | 1.00     | 0.3        | 50.0     | 8.0       |
| T9     | 1.00     | 0.5        | 16.6     | 24.0      |
thus resulting in a fine needle-shaped microstructure. Figure 7b depicts the fused zone, a large percentage of $\alpha'$ martensite, in addition to the secondary $\alpha$ acicular phase. The formation of $\alpha'$ occurred because the $\beta$ phase did not have enough time to precipitate $\alpha$; that is, it was possible to transform the network parameter; however, the $\beta$ composition remained almost unchanged, as a consequence: $\alpha$ and $\alpha'$ phases were formed through this transformation of the lattice parameter [45].

The HAZ, close to the fused zone, was affected by a greater amount of thermal energy during the process, in which the temperature varied between 980–995 °C (transition temperature $\beta$), and 1605 °C (solidus temperature). Therefore, the HAZ observed close to the FZ consisted mainly of the acicular martensitic phase $\alpha$ and a small amount of acicular $\alpha$, microstructures that are similar to that of the FZ [46, 47].

The volumetric fraction was measured using imaging software (ImageJ®) on the SEM images depicted in Fig. 6. In each welding condition, different values of volumetric fraction of the $\alpha$ phase in FZ region were found, for example, 13%, 18%, 32%, and 18% for samples T5, T6, T7, and T8, respectively. Thus, the volumetric fraction of $\beta$ in this region increased as it was subjected to higher values of heat input.

### 3.3 Formation of intermetallic compounds

Through the results so far, and in accordance with other studies, it is clear that the characteristics, size, distribution,
Fig. 7 a Microstructure of the base metal—titanium alloy (Ti6Al4V)—original increase of 500 ×. b Microstructure of the fused zone of the titanium alloy (Ti6Al4V), showing the α grain contour, original magnification 500 × magnification.

Fig. 8 Regions with intermetallic compounds: Face (left) and root (right) of welded joints, a sample T2, b sample T6, and c sample T7.
morphology of the phases formed in the laser welding process of the Al/Ti joint is administered by the coefficient of diffusion, welding energy, beam offset, and distance to the Al/Ti [48] interface. The diffusion coefficient of Ti in Al is $2.15 \times 10^{-8} \text{ m}^2/\text{s}$ [49], and it was observed that the phenomenon can occur at a relatively low temperature, around $600 \, ^\circ\text{C}$ [50].

The formation of intermetallic compounds was identified in the darkest gray regions, which can be seen in Figs. 8 and 9. In these places, either the aluminum alloy or the titanium alloy melted, or a composition of a reactive layer at the interface between Al/Ti occurred. It was identified in the images of the top, as well as the root, that the reactive layer of the interface is not homogeneous and tends to be curved. This behavior is related to the heat input and the interaction between the aluminum and titanium alloys in their liquid state.

Likewise, in the center of the welded joint shown in Fig. 9b (sample T9), a reaction layer was formed at the interface, in addition to the presence of the titanium lamellae that diffused in the aluminum where the temperature of recrystallization of aluminum was higher. The direction of growth of the lamellae followed that of the heat flow.

3.4 EDS analysis

The distribution of Al, Ti, V, and Si elements along the course depicted at Fig. 10a can be seen. Figure 10b shows the results

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Fig. 9 Regions with intermetallic compounds: a face (left) and root (right) of the welded joint, sample T8. b Center of the joint, sample T9 and c face (left) and root (right) of the welded joint, sample T9.
obtained by the EDS analysis across the Al/Ti welded interface.

Note the reduction of Al content as the line approaches to titanium alloy from the starting point of the analysis until 20 μm, where a peak of the Ti content occurs. It was identified that this region corresponds to approximately 0.5-μm layer. The presence of IMC could be assumed at this location. Furthermore, Table 9 shows the probable IMCs at each EDS analysis site measured at selected locations marked as Spectron of spots identified at Fig. 11 according to the chemical compositions presented in Tables 5, 6, 7, 8, to 9.

Considering the Ti-Al phase diagram (Fig. 12), the formation of the Ti₃Al (α₂) phase occurs in a composition range of 22 to 39% at. Al (14 to 19% wt). The phase is stable close to 1090 °C for Ti-25 Al (25%) stoichiometric composition and has a hexagonal DO₁₉ crystalline structure. The TiAl phase (γ titanium aluminide) has an aluminum composition between 49 and 66% at, with a L₁₀ tetragonal centered face structure. The phase remains in order until its melting point (1440 °C) [52].

### 3.5 Tensile test

The results of the tensile tests, for the welding conditions adopted, are presented in Figs. 13, 14, 15, 16, and 17. In Fig. 13, we observe the average values of the maximum stress with the respective standard deviations for samples T1, T2, T5, T6, T7, T8, and T9.

Note that conditions T6 and T9 exhibited the highest values of (tensile strength values shown in Fig. 14 was compared with the values of mechanical properties depicted in Table 2). In addition, as could be seen in Figs. 14, 15, 16, and 17, the total elongation (as a measure of the strain threshold) corresponded to conditions T6 and T9, being almost 6% of the initial length, as we can see in Fig. 16 and Fig. 17.

Furthermore, Fig. 18 shows the behavior of the tensile strength as a function of the welding energy, considering an offset of 0.5 mm and 0.3 mm. In both cases, it is noted that higher values of welding energy and offset tend to contribute to the increase in tensile strength, with a slight decrease in tension to 14 J/mm.

### 3.6 Hardness Test

The results of the hardness test on the dissimilar joint are presented in Figs. 19, 20, and 21. The mapping of indentations can identify regions made of fragile phases, resulting from variations in the microstructure that forms during the laser welding process. In order to better analyze the hardness map, the side of the aluminum alloy and the side of the titanium alloy next to the junction interface will be discussed separately.
3.6.1 Aluminum alloy side

Note that for the aluminum alloy side, in almost all conditions that were presented, the hardness in the region of the fused zone tends to increase as it moves away from the interface of the junction, even without changes in morphology and grain size. This is because the main hardening elements of the Al-Mg-Si alloy are Mg and Si. Mg evaporates during the welding process, due to its low melting point (650 °C) [53]. Approximately 0.3 mm from the interface (in the case of sample 6), the loss of alloy elements such as Mg is reduced, as the energy received from the laser beam attenuates, resulting in increased hardness. Such behavior does not occur on the map corresponding to sample 8, where there is a discontinuity along the entire Al/Ti interface, in addition to the FZ with less than 0.05 mm in length. Sample 9 shows greater homogenization of the hardness values.

### Table 5
Chemical composition measured at selected locations marked as spots 1 to 6, sample T6

| Position | Al  | Ti       | V          | Si          |
|----------|-----|----------|-------------|-------------|
| Spot 1   | 71.28/81.10 | 25.83/16.55 | 1.67/1.00 | 1.23/1.34   |
| Spot 2   | 40.97/55.19 | 55.7/42.26 | 3.02/2.15 | 0.31/0.40   |
| Spot 3   | 28/40.87   | 68.9/56.62 | 2.88/2.23 | 0.20/0.28   |
| Spot 4   | 23.62/35.46 | 73.16/61.86 | 3.02/2.40 | 0.20/0.28   |
| Spot 5   | 36.43/50.38 | 60.38/47.04 | 2.78/2.03 | 0.41/0.55   |
| Spot 6   | 25.41/37.74 | 70.18/58.70 | 4.25/3.34 | 0.16/0.22   |

3.6.2 Ti alloy side

The region analyzed for the titanium side covered the junction interface, the molten zone (FZ), and a part of the heat-affected zone (HAZ), except in the case of samples 6 and 9 that did not

### Table 6
EDS analysis at the places indicated as spots 1 and 2, intermetallic layer, sample T7

| Position | Al    | Ti    | Mg           | Si          |
|----------|-------|-------|--------------|-------------|
| Spot 1   | 44.28/58.29 | 54.98/40.77 | 0            | 0.74/0.93   |
| Spot 2   | 74.69/83.10 | 23.32/14.62 | 0.99/1.22   | 0.99/1.06   |
present significant HAZ near the interface. It is observed in the indentation map for the side of the titanium alloy that the gradient of the hardness values is larger and more heterogeneous, comparing them with the side of the Al.

In addition, the presence of small sites of high hardness, close to the junction interface, indicates the formation of intermetallic compounds. This fact is clearly perceived at approximately 0.05 mm from the junction interface for sample 6 (top of the weld) and 0.2 mm for sample 9 (root of the weld).

The highest hardness values in the molten zone occur due to the rapid cooling rate characteristic of the process, which results in a microstructure, consisting mainly of $\alpha'$ martensite.

On the other hand, the HAZ is formed by the $\alpha/\beta$ phases that have less hardness. In general, the different hardness values occur due to the offset and welding energy used. The table below shows the points with the highest hardness values for each condition analyzed with the respective positions on the welded joint.

### Table 7 Chemical composition measured at selected locations marked as spot 1 and spot 2, sample T5

| Position | Al    | Ti     | V       | Si    | Spot 1 | Spot 2 |
|----------|-------|--------|---------|-------|--------|--------|
|          | 44.54/58.74 | 51.97/38.60 | 3.10/2.17 | 0.39/0.49 |
|          | 37.24/51.30  | 59.65/46.28  | 2.86/2.09  | 0.25/0.33  |

### Table 8 Chemical composition measured at selected locations marked as spot 1 to spot 6, sample T8

| Position | Al | Ti | Mg | Si | F | V |
|----------|----|----|----|----|---|---|
| Spot 1   | 94.43/92.27 | 0.00 | 1.06/1.15 | 0.00 | 4.51/6.28 | 0.00 |
| Spot 2   | 7.45/12.53  | 87.41/82.88 | 0.00 | 0.00 | 5.14/4.59 |
| Spot 3   | 38.07/52.14 | 58.38/45.04 | 0.00 | 0.41/0.54 | 0.00 | 3.15/2.28 |
| Spot 4   | 26.64/39.21 | 69.71/57.81 | 0.00 | 0.21/0.30 | 0.00 | 3.44/2.68 |
| Spot 5   | 28.08/40.98 | 68.09/55.96 | 0.00 | 0.16/0.23 | 0.00 | 3.66/2.83 |
| Spot 6   | 8.33/13.92  | 86.85/81.78 | 0.00 | 0.04/0.06 | 0.00 | 4.79/4.24 |

The hardness values for the junction interface have been significantly changed as a function of the offset and the heat input. For the same offset value and higher heat input value, there was an increase in the hardness of the interface (see Table 10). An example is sample 9, compared to sample 1, with almost twice the welding energy, the hardness value increased by 100 HV0.2.

### 4 Conclusions

The experiments carried out and the results obtained in this work aimed to investigate and optimize the weldability of the dissimilar joint AA6013 and Ti6Al4V obtained by the laser welding process. Therefore, based on the materials and methods used in this work, it can be concluded that:

- The hardness values for the junction interface have been significantly changed as a function of the offset and the heat input.
- The presence of small sites of high hardness close to the junction interface indicates the formation of intermetallic compounds.
- The highest hardness values in the molten zone occur due to the rapid cooling rate characteristic of the process, which results in a microstructure, consisting mainly of $\alpha'$ martensite.
- The HAZ is formed by the $\alpha/\beta$ phases that have less hardness.

### Table 9 Identification of the intermetallic compounds

| Point | Sample 6 | Sample 7 | Sample 5 | Sample 8 |
|-------|----------|----------|----------|----------|
| 1     | Ti$_5$Al | Ti$_2$Al$_3$ | Ti$_2$Al$_3$ | MgF$_6$ |
| 2     | Ti$_5$Al$_4$ | Ti$_5$Al | TiAl | N.F.I* |
| 3     | TiAl | TiAl | TiAl | TiAl |
| 4     | Ti$_5$Al$_3$ | Ti$_5$Al$_2$ | TiAl | TiAl |
| 5     | TiAl | TiAl | TiAl | TiAl |
| 6     | N.F.I* | N.F.I* | N.F.I* | N.F.I* |

*N.F.I* It did not form intermetallic

![Fig. 12 Phase diagram of the Ti-Al system](image)
Welding conditions T6 (HI = 12.4 J/mm, Ws = 33.3 mm/s, offset = 0.5 mm), T7 (HI = 9.6 J/mm, Ws = 50.0 mm/s, offset = 0.3 mm), and T9 (HI = 24.0 J/mm, Ws = 16.6 mm/s, and offset = 0.5 mm) did not show significant discontinuities in the weld bead, thus meeting the acceptance criteria of the AWS D17.1 standard.

Fig. 13 Average value of maximum tensile stress for each welded condition

Fig. 14 Stress-strain curve—sample T1

Fig. 15 Stress-strain curve—sample T5

Fig. 16 Stress-strain curve—sample T6

Fig. 17 Stress-strain curve—sample T9

1. Welding conditions T6 (HI = 12.4 J/mm, Ws = 33.3 mm/s, offset = 0.5 mm), T7 (HI = 9.6 J/mm, Ws = 50.0 mm/s, offset = 0.3 mm), and T9 (HI = 24.0 J/mm, Ws = 16.6 mm/s, and offset = 0.5 mm) did not show significant discontinuities in the weld bead, thus meeting the acceptance criteria of the AWS D17.1 standard.

Table 10 Higher hardness values, beam offset, and heat input (HI)

| Sample | Higher hardness value (HV0.2) | Location | Offset (mm)/HI (J/mm) |
|--------|-------------------------------|----------|-----------------------|
| 1      | 442                           | ~0.2 mm from the interface near to the root | 0.5/14.4 |
| 5      | 506                           | ~0.15 mm from the interface near to the root | 0.3/14.4 |
| 6      | 520                           | ~0.05 mm from the interface near to the root | 0.3/12 |
| 7      | 589                           | ~0.2 mm from the interface near to the root | 0.3/9.6 |
| 8      | 506                           | ~0.2 mm from the interface near to the root | 0.3/8 |
| 9      | 570                           | ~0.02 mm from the interface near to the root | 0.5/24 |
2. The microstructure on the side of the aluminum alloy was characterized by the presence of α-Al grains and dispersed precipitates inside the grains. Between the coarse-grained region of the heat-affected zone (HAZ) and the melted zone, dendritic cellular grains were formed. In the HAZ, next to the region that was partially...
fused, elongated grains with epitaxial growth were formed, as well as equiaxial grains. Columnar grains grew epitaxially from the melting line, while equiaxial grains were formed in the liquid phase in the solidification direction. In addition, the average grain size for T1 condition was 83% larger than the average grain size in T2. Likewise, the welding energy values associated with the offset also interfered with the dissolution and volumetric fraction of the precipitates. The molten zone region for the titanium alloy exhibited the $\alpha'$ martensite phase, in addition to the secondary acicular phase. In the HAZ, a smaller volumetric fraction of $\alpha'$ was formed as it moved away from the molten zone. In the region close to the Al/Ti interface, phases $\alpha$ and $\beta$ and a small amount of $\alpha'$ were also identified. The volumetric fractions of $\alpha$ were 13, 18, 32, and 18% for samples T5, T6, T7, and T8, respectively. The volumetric fraction of $\beta$ in this region increased as it was subjected to higher values of heat input.

3. Conditions T6 and T9 showed the highest values of tensile strength compared to the other experiments, around 60 and 50% (respectively) of the resistance of the aluminum alloy. The presence of the reactive layer between the titanium alloy and the aluminum alloy along the joint, the curvilinear shape on the face and root, as well as the layer thickness of intermetallic compounds acted as agents, promoting coalescence. The main intermetallic compounds identified in these cases were TiAl, Ti$_3$Al$_4$, Ti$_2$Al$_3$, and Ti$_5$Al$_3$.

4. The hardness values for the molten zone on the Al side were not strongly affected by changes in offset and welding energy. The titanium alloy side, on the other hand, showed more heterogeneous hardness values, in addition to forming sites with high hardness values close to the junction interface, caused by the presence of intermetallic compounds. In addition, the hardness at the joint interface has been significantly changed due to the offset.
and the welding energy. For the same offset value and higher welding energy value, there was an increase in the hardness of the interface. An example is sample 9, in comparison to sample 1, with almost twice the welding energy, got a hardness increase value of 100 HV0.2.

5. In conclusion, the extension of the HAZ, the volumetric fraction of the phases, and intermetallic compounds are associated with the value of the offset and the welding energy used in each condition of the experiments. Likewise, the appearance of the weld bead is closely related to the action that the molten metal (in this case, aluminum) had on wetting the titanium alloy interface.

Authors’ contributions Anderson Ribeiro, Milton Fernandes, and Jorge Abdalla conceived and designed the laser welding experiments. Anderson Ribeiro, Milton Fernandes, and Rafael de Siqueira performed the experiments. Anderson Ribeiro and Jorge Abdalla analyzed the data. Anderson Ribeiro and Rafael Giorjão discussed the results. Anderson Ribeiro wrote the paper. Anderson Ribeiro and Rafael Giorjão revised the manuscript.

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Code availability Not applicable.

Declarations

Ethical approval Not applicable.

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Conflicts of interest The authors declare no conflict of interest.

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