Experimental and numerical study on microstructure and mechanical properties for laser welding-brazing of TC4 Titanium alloy and 304 stainless steel with Cu-base filler metal

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A B S T R A C T
Laser welding-brazing of TC4 Titanium (Ti) alloy to 304 stainless steel (SS) has been applied using 38Zn-61Cu alloy as filler metal. Microstructures of the joints were studied using scanning electron microscopy (SEM), energy dispersive spectroscopy (EDS) and X-ray diffraction (XRD). Mechanical properties of the joints were evaluated by performing tensile tests. The temperature field and stress field distributed in laser welding based on SS-Ti alloy joint were dynamically simulated using the ANSYS in this study. A new welding process for SS-Ti alloy joint was introduced on the basis of the controlling the formation of Ti-Fe intermetallics in the joint. One process was one pass welding involving creation of a joint with one fusion weld and one brazed weld separated by remaining unmelted SS. When laser beam on the SS side was 1.5 mm, SS would not be completely melted in joint. Through heat conduction, the filler metal melted occurred at the SS-Ti alloy interface. A brazed weld was formed at the SS-Ti alloy interface with the main microstructure of (Fe, Zn)+FeZn2, β-CuZn and β-CuZn+Ti2Zn. The joint fractured at the brazed weld with the maximum tensile strength of 210 MPa. By comparing the simulation results with the corresponding experimental findings, the validity of the numerical model is confirmed.

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1. Introduction

Recently, aerospace and nuclear industries have a strong demand of dissimilar Ti alloy to stainless steel (SS) for applica-

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emergence of continuously distributed TiFe, TiFe₂. Hardness of the weld metal was in the range of HV740-HV1324 [4-6]. The presence of these brittle phases decreases the strength and plasticity of the joint. In this case, complex intermetallic compounds form between Ti, Fe, Cr and Ni which made the weld even more brittle [7,8]. Consequently, resulting welds are very brittle and do easily crack. Moreover, the joint excessive distortion and residual stresses because of the significant differences in their physical properties, such as coefficient of thermal expansion and melting point [9]. Current research suggests that conventional fusion welding would result in the formation of a thick brittle intermetallics layer and an accumulation of residual stress at the joint [10]. It has been acknowledged that direct fusion welding methods are not feasible for the joining of Ti alloy-SS because of their metallurgical incompatibility.

Currently, indirect joining is generally realized by adding an interlayer such as Cu, Ni, or Ag to prevent atomic diffusion between Ti and Fe, Cr, or Ni [11-14]. As those materials form intermetallic phases with Ti and Fe, the strength of such welds depends on brittleness of TiₓMᵧ (M-metal of interlayer) comparing with TiFe and of spatial distribution of intermetallics in the joints [7]. However, as long as the interlayer was completely melted, Ti and Fe elements would mix and react in the weld pool and Ti-Fe intermetals would be produced in the weld. Inversely, pressure welding can eliminate the problems in the direct fusion welding because the base metals remain in the solid state during joining and many successful examples have been reported [15-17]. S. Kundu et al. [15] have studied Ti-6Al-4 V and micro-duplex stainless steel was diffusion bonded in vacuum. Effect of bonding temperature and time on the strength properties at room temperature were evaluated. Shear strength of 397.5 MPa along with 6.5% elongation was obtained for the diffusion couple processed at 850 °C for 90 min. Fazel-Najafabadi et al. [16] achieved friction stir welding parameters were adjusted in order to achieve defect-free dissimilar lap joint of CP-Ti to 304 stainless steel. Joint shear strength was measured; a maximum failure load of 73% of that of CP-Ti was achieved. Mousavi and Sartangi [17] have arrived at a suitable parametric window both analytically and experimentally for explosive welding of CP Ti-SS 304. They have concluded that at low loads formation of intermetallics could be totally avoided. However, the service conditions may make particular processes unsuitable. The diffusion process needs a long time to implement in general. Furthermore, the required joint geometry can make friction welding and explosive welding difficult to apply [18]. Moreover, brazed welding can also eliminate the Ti-Fe intermetallics in the joints due to the base materials remain in the solid state during welding [19].

In fact, butt welding of Ti alloy to steel was convenient in technology. As a non-contact fusion joining technique with high efficiency and flexibility, laser welding has made great achievements in the joining of hard to weld materials and dissimilar metals [20]. Laser welding was particularly suitable for welding of materials with high thermal diffusivity and conductivity, crack-sensitive, different melting points [21]. Dürr et al. [22] improved weldability in laser welding of refractory metals and dissimilar metals. Chen et al. [23] investigated the effect of laser-beam offsetting on microstructural characteristics and fracture behavior of the joint. When the laser beam is offset by 0 mm toward the Ti alloy, the joints fracture spontaneously subsequent to welding. When the laser beam was offset to 0.6 mm toward the Ti alloy side, the tensile strength of the joint was only 24.75 MPa. It was found that when the laser beam was offset to 0.6 mm toward the stainless steel side, it results in a more durable joint. The tensile strength of the joint was 150 MPa. In this case, the shift of heat source to stainless steel allowed to reduce the melting of the Ti alloy and subsequently reduced the amount of Ti-Fe intermetallics, but liquid-states mixing between Ti alloy and steel was not suppressed fully. It was shown that the formation of a thin layer of the Ti-Fe intermetals takes place for any beam offset. Therefore, the coarse and unordered Ti-Fe intermetallics in the joint are inevitably formed if the base metals was fully melted. Thus, base metals liquid-state mixing of Ti alloy-SS joint should be avoided during welding. Based on the above analysis, two welding mechanisms (fusion welding and brazing) are combined to avoid melting and liquid mixing of the base metals during welding. Using laser as welding heat source and Cu-based fillers as interlayer material, the formation of Ti-Fe intermetallics in Ti alloy-SS joint was avoided by laser welding-brazing, the brittleness of joint is reduced and the properties of joint are improved.

In view of the above analysis, use of the laser welding-brazing was proposed in this paper with the main objective to avoid mixing of Ti alloy and SS. The welding process was set up to ensure that the SS was partly melted. The melted SS formed a fusion weld. Meanwhile, a brazed weld was formed at the interface between unmelted SS and Ti alloy. In this way, a peculiar joint was acquired and Ti-Fe intermetallics can be completely avoided in the joint since the unmelted SS acted as a diffusion barrier. The relation between joint microstructures, mechanical properties and fracture modes was discussed in detail.

2. Experimental procedure

2.1. Materials

The base materials used in this experiment were TC4 Ti alloy and 304 stainless steel. Their chemical compositions and physical properties are given in Tables 1-3. It can be seen that there are large differences in thermal conductivity and linear expansion coefficient between the two base materials, which would lead to large temperature gradient and thermal stress in the joint during welding process. The base materials were machined into 100 mm × 80 mm × 1 mm plates, and then
cleaned with acetone before welding. The filler metal used was 0.2 mm plate of (melting point 820 ºC) Cu-base filler metal (61.2 wt.% Cu, 37.2 wt.% Zn, 0.28 wt.% Si and 0.89 wt.% Sn). Before welding, the specimens were mechanically and chemically cleaned. The gap between the edges of the Ti alloy and SS was very important to adequate heat transfer and prevent porosity formation. The specimens are clamped each other tightly in order to get the minimum gap formation between the edges.

2.2. Welding method

CW laser was used with average power of 1.20 kW, wavelength of 1080 nm and beam spot diameter of 0.1 mm. A schematic diagram of the welding procedure is shown in Fig. 1a and Fig. 1b. In order to ensure that SS was not completely melted, the laser beam was focused on the SS plate 1.5 mm away from the SS-filler metal interface. Laser welding-brazing involving creation of a joint with two weld zones separated by remaining unmelted SS. Laser welding-brazing involving creation of a joint with two weld zones separated by remaining unmelted SS. The welding parameters were: laser beam power of 624 W, defocusing distance of +5 mm, welding speed of 600 mm/min.. Argon gas with the purity of 99.99% was applied as a shielding gas with total flow of 20 L/min at top of the joint.

2.3. Characterization methods

The cross sections of joints were polished and etched in the reagent with 2 mL concentrated HNO₃ and 6 mL concentrated HF. The microstructure of joints were studied by optical microscopy (Scope Axio ZEISS), scanning electron microscope SEM (S-3400) with fast energy dispersion spectrum EDS analyzer and selected area XRD (X’Pert3 Powder) analysis. Vickers microhardness tests for the weld carried out with a 10 s load time and a 200 g load. Tensile strength of the joints was measured by using universal testing machine (MTS Insight 10 kN) with cross head speed of 2 mm/min. WRN-191 K sheathed thermocouple (measuring range −250–1350 ºC) as the temperature sensor.

2.4. Numerical model and simulation

The software ANSYS has been broadly employed to solve the problem of computational fluid dynamics. The commercially available finite-element software ANSYS was employed to calculate the temperature and stress distributions. Generally, the heat source model was considered an important aspect of the welding thermal analysis, especially when the joint was thicker [24]. Energy input was simulated, so that the weld dimensions and shape complied with those prepared experimentally with known laser welding parameters. A combined heat source model of gaussian surface heat source and double ellipsoid body heat source was applied. Since the temperature gradient was very steep, the melting pool was finely meshed to obtain reliable results of finite element method. The grid size (0.05 mm × 0.05 mm × 0.05 mm) was dense in and near the weld to ensure a more accurate simulation.

Its controlling equation of heat source model is written as:

\[ q_1(x, y, t) = \frac{3\eta_1Q}{\pi r^2} e^{-\frac{30^2}{r^2 + (x - y)^2} + \frac{30^2}{r^2 + (x + y)^2}} \] (1)

Where: Q is the total power; \( \eta_1 \) is the ratio of the Gaussian surface heat source to the total heat source power; \( \eta_1 \) is the actual effective power ratio coefficient; \( r \) is the effective heating radius; \( v \) is the welding speed, \( t \) is the welding time; and \( r \) is the delay time.

3. Results and discussion

3.1. Macro-characteristics

The optical microscopy image of the cross section of the joint is shown in Fig. 2a. The joint can fall into three parts: the fusion weld formed at the SS side, unmelted SS and the brazed weld formed at the SS-Ti alloy interface. The fusion weld did not form Ti-Fe intermetallics due to the presence of unmelted SS. The average width of fusion weld, unmelted SS and brazed weld was 1.85 mm, 0.35 mm and 0.14 mm, respectively. Moreover, the local heating of the SS side caused uneven volume
expansion and thermal stress was produced, which helped to obtain an intimate contact between the SS, Cu-based fillers and Ti alloy surface. However, the filler melts instantaneously during welding, and molten filler was squeezed from the Ti alloy-SS interface under the stress to form a filler droplet on the surface. Thus, braze welding occurred, and a brazed weld was produced at SS-Ti alloy interface. The brazed weld becomes black by metallographic etching. Fig.2b presents the optical microscopy image before corrosion of the brazed weld. It does not present such defects as pores and macro-cracks. The unmelted SS was left in the joint after welding with the main objective to avoid mixing of Ti and Fe elements so that the formation of Ti-Fe intermetallics was expected to be eliminated in the joint. Additionally, the unmelted SS was beneficial to relieve and accommodate the thermal stress in the SS-Ti alloy joint, which could help to improve the mechanical properties of the joints.

It should be noted that precise control of the laser spot position is crucial to obtain a sound joint. If the laser spot is far away from the SS-Ti alloy interface, brazing process at the SS-Ti alloy interface cannot take place. If the laser spot is draw near to the SS interface, the SS is beginning to melt in the joint, thereby amount of brittle Ti-Fe intermetallics is greatly increased in the joint and cannot be realized the effective combination between dissimilar materials of SS and Ti alloy.

Fig. 3 presents the physical model formation of the joint. When the laser beam was fixed on the side of the SS plate with some distance from SS interface, the welding pool with keyhole would be produced inside the SS plate, as seen in Fig. 3a and Fig. 3b. Once the keyhole was generated, massive heat would be absorbed. The SS as the keyhole boundary melted sharply due to the high absorption of the laser inside the keyhole. Because of the laser beam was fixed on the SS side with a farther distance from SS-filler metal interface, SS side was not completely melted, as seen in Fig. 3b. Thus, the temperature of unmelted SS increased rapidly as the keyhole boundary due to the high absorption of the laser inside the keyhole. Meanwhile, the unmelted part of SS due to its high thermal conductivity of absorbing a significant amount of heat from the welding pool and transferring it to the Ti alloy side, as seen in Fig. 3c. The simulation results of the temperature field in laser welding of Ti alloy and SS were studied by Hu Xiaohong et al. [25]. Although the energy density was concentrated, the temperature was higher in the area of 1 mm from the center of the heat source. Within the range of 1–2 mm from the laser incident point, there is a temperature gradient of 1522–400 °C. Hence, the unmelted SS had a high temperature which was high enough to promote filler metal melting at SS-Ti alloy interface, as seen in Fig. 3d. Because the thermal conductivity of SS was significantly higher than that of Ti alloy, the solidification of the weld pool and liquid filler metal in the joint will start from the SS side, leading to the formation of the composite joint between the fusion weld and brazed weld, as seen in Fig. 3e and f.

3.2. Thermal cycling test of SS-Ti alloy interface during welding

To better reflect the formation of the brazed weld at SS-Ti alloy interface, the thermal cycling test was performed on the SS-Ti alloy interface. Due to the limitation of plate thickness and clamping device, the welding heat cycle is difficult to test through drilling in the middle or bottom of the test plate. Moreover, the heat of the SS-Ti alloy interface was derived from the welding pool by using unmelted SS. The thermal cycling of the SS-Ti alloy interface can be accurately measured only under the contact between thermocouple and unmelted SS, Ti alloy during welding. To simplify the test and make the test more precise, this paper studied the heat cycle test from the side slot of the test plate. The structure was provided with opposite grooves on the Ti alloy side faces, and the interior of a sealing groove formed by the groove was provided with a thermocouple, as seen in Fig. 4a. This greatly simplifies the internal installation of thermocouples in the joint. The thermal cycle curve obtained from SS-Ti alloy interface during welding is shown in Fig. 4b. It is suggested that the peak temperatures of brazing interface (SS-filler metal and filler metal-Ti alloy) were 1094 °C and 905 °C respectively, which was above the melting point of filler metal but below the melting point of SS and Ti alloy. This meets the temperature requirement for braze welding. Thus, filler metal melting, and a brazed weld was produced at SS-Ti alloy interface. Moreover, the heating rate and cooling rate of the SS-Ti alloy interface were very fast. This is primarily because the laser welding has faster heating and cooling rate. Besides, the high thermal conductivity of SS also increases the heating and cooling rates of the SS-Ti alloy interface. Given that the SS-Ti alloy interface had faster heating and cooling rate, the holding time at high temperature was

Fig. 2 – Macroscopic feature of the joint: (a) optical image of the cross section of the joint; (b) optical image before corrosion of the SS-Ti alloy interface.
short, the brazing process at the SS-Ti alloy interface was very fast.

3.3. The simulation of laser welding temperature field

The contours of the temperature field frontal view of the joint are shown in Fig. 5a. Because the welding process was stable when the laser beam moved to the middle of the Y coordinate, the y = 17 mm cross section was selected to investigate the temperature distribution around the fusion zone. Under the condition of moving thermal resource, the temperature distribution of workpiece changes quickly with the variety of time and space [26]. It is suggested from Fig. 5a that the distribution of temperature field varies as the laser source and the weld pool move along with the laser source. The fusion zone has clearly extended to the SS side. The melt course of SS can be clearly observed in Fig. 5a. The SS was not melted thoroughly due to the laser beam was focused on the SS plate.

Fig. 3 – Physical model formation of the joint: (a) laser beam was focused on the SS plate; (b) formation of welding pool on SS side; (c) heat was transferred from welding pool to the SS-Ti alloy interface; (d) melting of the filler metal on SS-Ti alloy interface; (e) formation of fusion zone on SS side; (f) formation of brazed weld on SS-Ti alloy interface.

Fig. 4 – Temperature measurement test of SS-Ti alloy interface during welding: (a) thermocouple distribution; (b) thermal cycle curve of test points.
1.5 mm away from the SS-filler metal interface. Meanwhile, the unmelted part of SS due to its high thermal conductivity of absorbing a significant amount of heat from the welding pool and transferring it to the Ti alloy side. Hence, the filler metal at SS-Ti alloy interface had a high temperature. The contours of the temperature field cross section of the joint during welding are shown in Fig. 5b. In general, the color contours in the simulation images provide a convenient means of predicting the extent of the HAZ with in the weldment [27]. It is noteworthy that in the simulation images, the red region represents the temperature above 2700 °C (i.e. the melting point of 304 SS 1454 °C) and thus corresponds to the fusion zone of the joint. The peak temperature of weld pool was nearly 3300 °C. At this temperature, the SS can be melted under the action of laser heat, and the width of the molten SS was nearly 2.05 mm. For the filler metal at SS-Ti alloy interface, there is a certain delay in time through the heat conduction of unmelted SS. That is, the temperature of filler metal at SS-Ti alloy interface reaches a maximum after a few seconds of laser passing. The contours of the temperature field cross section (0.3 seconds later) during welding are shown in Fig. 5c. The temperature gradient of unmelted SS was about 1400–1119 °C, and the width of unmelted SS was nearly 0.31 mm. The filler metal of SS-Ti alloy interface are under the high temperature shock though there is no heat source in SS-Ti alloy interface. The temperature gradient of filler metal was about 1110–945 °C. The results show that the simulated results were well consistent with the experimental measurements. This shows that the entire filler metal was at a relatively high temperature. This meets the temperature requirement for braze welding. Thus, filler metal melting, and a brazed weld was produced at SS-Ti alloy interface. Thermal conductivity of SS is almost two times as Ti alloy to heat evenly to the external heat sink. The heat conduction of unmelted SS becomes the sole mechanism after this time in this region.

3.4. The simulation of laser welding stress field

The simulation of stress field in transverse direction with the joint during welding is shown in Fig. 6. It is suggested from this figure that large residual stress were formed in the fusion zone and its vicinity. High intensity laser beam melted and partially evaporated the SS during welding. The high temperature gradient formed in the heating and cooling periods develops high thermal stresses in the fusion zone. Once the cooling period ends, the residual stress in the fusion zone will be formed [28]. It is suggested from Fig. 6 that the residual stress distribution of SS-Ti alloy interface was not symmetrical, and stress level of SS side is significantly higher than the filler metal and Ti alloy. This is primarily because the laser beam was focused on the SS side, which causes the SS to melt in a large amount. In such case, molten metal of SS side was free to expand in fusion zone. The rapidly transmitted stress in unmelted SS soon hit the SS-Ti alloy interface. Accordingly, a certain amount of stress will be formed in the SS-Ti alloy interface. Moreover, the SS, filler metal and Ti alloy have larger difference in linear expansion coefficient, and linear expansion coefficient of SS is almost five times as filler metal. Thus, the residual stress had low values in the SS-Ti alloy interface due to the presence of filler metal. By comparison, the filler metal of SS-Ti alloy interface is easy to release stress, and improve the mechanical properties of the joint.

3.5. Microstructure analysis

The optical image of the fusion weld is shown in Fig. 7a–c, and no defects were observed in it. SEM image of the fusion weld is shown in Fig. 7d. The fusion weld mainly consists of columnar crystal. As seen in Fig. 7b, the fusion line of fusion weld has the characteristic of epitaxial growth, the cellular crystal is attached to the unmelted grain nucleation and growth, and its growth direction is approximately perpendicular to the fusion line. As the cellular crystal grows toward the weld centerline, the cellular crystal changes into the cellular dendrite crystal, and a small amount of dendritic crystals can be seen in the middle of the weld, as seen in Fig. 7c. The above changes in the weld crystal morphology are mainly related to the temperature gradient and constitutional supercooling of the welding pool. In the cooling stage of the welding pool, the formation of cellular crystals is favorable due to the large temperature gradient at the edge of the welding pool and the small constitutional supercooling zone. As the solidification process of the welding pool progresses, the temperature gradient decreases and constitutional supercooling interval increases, which promotes the transition of the cellular crystal to the cellular dendritic crystal.

The optical image of the brazed weld is shown in Fig. 8a, and no defects were observed in it. The white lamellar like
columnar particles near the SS side was very obvious, and a large amount of black coarse particles grow in the direction perpendicular to the Ti alloy side. It can be observed that, the brazed weld contained three zones marked as I, II and III sorted by their morphologies and colors. SEM image of the brazed weld is shown in Fig. 8b–d. The Fig. 8b–d correspond to the three zones in Fig. 8a, respectively. It can be seen from that the brazed weld can be divided approximately into three parts, namely the coarse grain structure zone near the SS side (position A), the fine grain zone in the centre of the reaction layer (position B), the column grain zone near the Ti alloy side (position C). The compositions of each zone (denoted by let-

### Table 4 – The chemical composition of each phase (at.%).

<table>
<thead>
<tr>
<th>Region</th>
<th>Composition%</th>
<th>Potential phases</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>Ti</td>
<td>Fe</td>
</tr>
<tr>
<td>A</td>
<td>56.0</td>
<td>44.0</td>
</tr>
<tr>
<td>B</td>
<td>59.3</td>
<td>22.8</td>
</tr>
<tr>
<td>C</td>
<td>30.5</td>
<td>25.1</td>
</tr>
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</table>
Cu-Zn-Ti was applied to these zones to measure the compositions of the reaction products and the results are listed in Table 4.

Based on the previous analysis, the microstructure of the brazed weld was mainly composed of molten Cu-based fillers. According to EDS analysis results, main composition in position A of zone I was of 56.0 wt.% of Cu and 44.0 wt.% of Zn. Therefore, the chemical composition of zone I was consistent with the Cu-based fillers. Based on the EDS analyses results and Cu-Zn binary phase diagram [29], the main microstructure of zone I was defined as β-CuZn phase. When the liquid phase is produced, the element diffusion occurs immediately between the base materials and liquid phase, and causes its component to deviate from the original component. Therefore, the liquid phase generation and the element diffusion occur simultaneously. Accompanied by the dissolution of solid phase into the liquid phase, Cu and Zn in the liquid phase would diffuse into SS and Ti alloy, which formed solid-phase reaction layer, and this reaction layer exists only in the smaller region of the interface. As shown in Fig. 8c and d, zone II and zone III were reaction layers formed by element diffusion. The composition at position B of zone II was 59.3 wt.% Fe and 22.8 wt.% Zn. Based on Fe-Zn binary phase diagram [30], the microstructure of zone II was defined as (Fe, Zn)+FeZn$_2$. The composition at position C of zone III was 30.5 wt.% Ti, 43.1 wt.% Zn and 25.1 wt.% Cu. Based on Cu-Zn-Ti ternary phase diagram [31], the microstructure of zone III was defined as β-CuZn+Ti$_2$Zn$_3$. Therefore, the main microstructures of brazed weld were (Fe, Zn)+FeZn$_2$, β-CuZn and β-CuZn+Ti$_2$Zn$_3$. According to EDS analysis results in Table 4, the content of Fe element in zone II is significantly higher than that Ti element in zone III. In another word, the larger amount of Fe element diffused into Ti alloy side than that of Ti element diffused into SS side. This is due to the higher diffusivity of Fe into Ti than Ti into Fe [32], that is, Fe has a faster diffusion rate than Ti in liquid phase. Thus, the Fe content in the liquid metal near the SS side diffuses farther.

Fig. 9 present the SEM-EDS plane analysis results from WZ$_2$. Based on the EDS line analysis in Fig. 9, Cu and Zn, as two main elements, were detected in the brazed weld, and content of Ti and Fe elements was very low. It can be seen from Fig. 9 that the content of Cu element in the brazed weld is significantly higher than that of Zn element. In order to confirm the elements distribution in the brazed weld, the SEM-EDS line analysis were carried out, as shown in Fig. 10. The line analysis started from Ti alloy side, passed through the brazed weld and ended in 304 SS side. Moreover, from the Ti alloy to the 304 SS, the content of Fe element increased rapidly while the content of Ti element decreases rapidly as a whole.

A schematic of formation process of brazed weld was presented in Fig. 11. When the laser beam was focused near the SS-Ti alloy interface, the temperature at the SS-Ti alloy interface was increased due to heat conduction of unmelted SS although it was not subjected to laser radiation. The high temperature of SS-Ti alloy interface promoted filler metal to melt and atomic interdiffusion, as shown in Fig. 11a and Fig. 11b.

Fig. 8 – Microstructures in the brazed weld of the joint: (a) optical image of the brazed weld; (b) SEM image of the zone I in Fig. 8a; (c) SEM image of the zone II in Fig. 8a; (d) SEM image of the zone III in Fig. 8a.
higher temperatures, the thermal energy supplied to the diffusing atoms permitted the atoms to overcome the activation energy barrier and more easily move to new lattice sites [33]. At this moment, the dissolution of Ti and Fe into the liquid phase occurred at solid-liquid interface under the high concentration gradient, which would change the composition of the liquid, as shown in Fig. 11c. Since the liquid existed for only a short time due to the laser welding with rapid heating and cooling, Ti and Fe in the liquid did not have enough time to spread evenly leading to concentration gradient in the liquid. In the subsequent cooling process, the liquid phase with different compositions would experience different reaction. So it can be concluded that at position A, β-CuZn solid solution was primarily precipitated from the liquid during cooling, as shown in Fig. 11d. When the temperature reduces, the (Fe, Zn) + Fe3Zn7 phase zone arises at SS side, the β-CuZn + Ti3Zn3 phase zone arises at Ti alloy side, as shown in Fig. 11e and Fig. 11f. After liquid phase solidified completely, the brazed weld was formed in the SS-Ti alloy interface. In this case, the microstructure of the brazed weld was mainly composed of Fe-Zn intermetallics were formed at SS side but Ti-Zn intermetallics was formed at the Ti alloy side. The β-CuZn + Ti3Zn3 phase columnar particles near the Ti alloy side was very obvious (Fig. 8a and Fig. 11f). Because the highest temperature gradient was produced along the direction perpendicular to the solid/liquid interface [34], gains grow rapidly along this direction, which induces formation of the coarse columnar crystal structure. At higher cooling rates directionality was observed in the growth of the brazed weld resulting in an oriented microstructure with the microstructure aligned in the direction of heat flow. Because of the formation of a directional dendritic structure introduces a number of anisotropy to the mechanical properties of the material which typically show significantly more robust mechanical properties along the dendrite growth direction [35]. Therefore, the brazed weld has high mechanical property, made by a laser beam offset toward the SS side.
Therefore, the laser SS-Ti reaction is used to increase the thickness of the intermetallic layer, and the growth of the reaction layer can be controlled by the atoms diffusion, in which the thickness of intermetallic layer is written in Eq. (2):

$$X = K_0 \exp(-Q/RT)\sqrt{t}$$  \hspace{1cm} (2)

In Eq. (2), $X$ denotes the thickness of intermetallic layer, $t$ is reaction time, $n$ is time factor (0.5), $K_0$ is constant, $R$ is the gas constant ($R = 8.314/(\text{mol} \cdot \text{K})$), $Q$ is the diffusion activation energy and $T$ is the reaction temperature. Eq. (2) suggests that both $t$ and $T$ can widen the intermetallic layer, and the contribution from $T$ is considerably larger than that from $t$. Therefore, the thickness of reaction layer at SS-Ti alloy interface increases with the increase of the reaction temperature. As mentioned before, the heating rate and cooling rate of the SS-Ti alloy interface were very high. This is mainly because the laser welding has a faster heating and cooling rate, and the holding time at high temperature was short. Therefore, the heating rate and cooling rate of the SS-Ti alloy interface were very fast by heat conduction of unmelted SS, and the reaction layer existed only in a smaller region of the SS-Ti alloy interface, which readily forms a narrower reaction zone at the SS-Ti alloy interface. By reasonably controlling laser welding parameters, the thickness of the reaction layer can be kept comparatively low to obtain a small amount of intermetallics, which contributes to the mechanical properties of the brazed weld.

3.6. Microhardness tests

As shown in Fig. 12, the microhardness distribution in the joint was non-uniform. TC4 Ti alloy has similar hardness to 304 SS. It can be seen from Fig. 12 that the microhardness distribution in the fusion weld was not uniform, approximately W type. The microhardness of the fusion weld was significantly reduced compared to the 304 SS. This is because the melting and re-solidification and crystallization of the SS base metal under the action of laser heat source during welding, which releases the internal stress and distortion energy of the base metal in the original cold working hardening state, which leads to the softening of the fusion weld and the decrease of the microhardness value here. The highest hardness of the fusion weld is located at the weld center, because the temperature of center of the welding pool is the highest during welding, where the concentration of the solute reaches a maximum, and the alloying elements by solid solution strengthened to increase the hardness. Moreover, the hardness of the brazed weld was very low compared to the fusion weld. The hardness of brazed weld was low because filler metal was simple metals. Therefore it can relatively deform easily to reduce the residual stresses in the inner of SS-Ti alloy joint.
Fig. 12 – Vickers microhardness measurements at semi-height of the joint (zero point situated in the center of the joint).

Fig. 13 – Tensile test results of the joint: (a) tensile test curve; (b) fracture location; (c) SEM image of fracture surface; (d) XRD analysis results of fracture surface.
3.7. Tensile tests and fracture analysis

The maximum tensile strength of the joint was about 210 MPa (Fig. 13a). The joint fractured in Ti alloy side of the brazed weld during tensile tests (Fig. 13b). Fig. 13c shows fracture surface of the joint exhibiting typical brittle characteristics. Moreover, as shown in Fig. 13d, XRD analyses of fracture surface detected β-CuZn and Ti₂Zn phases. This confirmed the presence of Ti-Zn intermetallics at fracture surfaces. It should be noted that there was no Ti-Fe intermetallics in the brazed weld. Reaction layer at Ti alloy side in brazed weld became the weak zone of the joint, which led to the failure in the tensile test.

4. Conclusions

With a laser beam offset of 1.5 mm for SS side of the joint, the unmelted SS was selected as an barrier to avoid mixing of the Ti alloy and SS. A brazed weld was formed at the SS-Ti alloy interface with the main microstructure of (Fe, Zn) + Fe₅Zn₇, β-CuZn and β-CuZn + Ti₂Zn₃. A great amount of liquid forms in the SS-Ti alloy interface during welding, and the thickness of brazed weld can reach hundreds of micrometres. The tensile resistance of the joint was determined by brazed weld. The maximum tensile strength of joint was 210 MPa.

By calculating temperature field and stress field, the structure integrity of joint in the welding was studied. The experimental and simulation results have suggested that the area enclosed by the temperature field contours in the brazed weld was less than that enclosed by the temperature field profiles in the fusion weld. Furthermore, the maximum value of residual stress generated by the brazed weld was much smaller than that caused by the fusion weld.

Conflict of interest

We declare that we do not have any commercial or associative interest that represents a conflict of interest in connection with the work submitted.

References

[27] HuaTeng L, ChunTe C, JiaLin W. Numerical and experimental investigation into effect of temperature field on sensitization of Alloy 690 butt welds fabricated by gas


