

Microstructural evolution and shear strength of brazing C103 and Ti–6Al–4V using Ti–20Cu–20Ni–20Zr (wt.%) filler metal

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Abstract

Microstructural evolution and shear strength of the brazed C103 and Ti–6Al–4V joint using the Ti–20Cu–20Ni–20Zr (wt.%) filler metal were studied. The shear strength of the joint brazed at 950 °C increases with the brazing time. However, the shear strength of the joint brazed at 1000 °C decreases with the increasing brazing time, because of the formation of coarse acicular Widmanstätten structure in the joint interface during brazing. In the brazed joint, the width of eutectoid and Widmanstätten regions increase with the increasing brazing time. On the contrary, the thickness of central brazed layer reduces with the increasing brazing time. Additionally, it was found that to obtain a high joint strength of approximately 300 MPa, both the central brazed layer and the coarse Widmanstätten structure should be avoided forming in the brazed joint.

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1. Introduction

Niobium is a refractory metal with high melting point of 2468 °C, relatively low density of 8.75 g/cm³ and good strength at high temperature [1,2,6]. Therefore, niobium and its alloys are widely applied for rocket applications. The C103 alloy is one of the most important niobium alloys to apply in liquid rocket thrust cones and aircraft turbine engine afterburners, which are commonly used in satellite launch system and navigation control thruster, due to its excellent welding properties and the elevated-temperature strength [1–3]. The nominal composition of C103 alloy in weight percent is 1%Ti, 10%Hf and the balance of Nb. Additionally, the Ti–6Al–4V alloy, an α - β Ti alloy which can be strengthened by solu-

tioning and aging heat treatments, is the most important and widely used titanium alloy [2,4,5]. The combination of high specific strength, good tensile strength at moderate temperature and excellent corrosion resistance made the Ti–6Al–4V alloy widely applied in the aerospace industry.

To date, much work has been done in the area of vacuum brazing titanium alloys to themselves [13–15], but very little has been published regarding the brazing of titanium to other metals, especially to the refractory metal. However, the necessity of joining C103 alloy to Ti–6Al–4V alloy would be encountered in practical production of liquid rocket [1,2,4–6]. Our previous studies indicated that Ti–15Cu–15Ni (wt.%) filler metal can be successfully used to braze Ti–6Al–4V and C103 alloys [17]. Nevertheless, the melting point of the Ti–15Cu–15Ni (wt.%) filler metal is about 960 °C, which is approximately the same as the β -transus temperature of Ti–6Al–4V alloy. In the case of brazing below β -transus temperature of the Ti–6Al–4V base metal, the original

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fine grain of the base metal can be completely preserved. On the contrary, in the case of brazing above the β -transus temperature of the Ti–6Al–4V base metal, the grains of the base metal would be coarsen and coarse Widmanstätten structure would be formed in the joint interface, deteriorating the mechanical properties of the brazed joint [4,5]. According to previous studies, for Ti-alloy joints, the formation of the coarse Widmanstätten structure causes the decreasing of the shear strength and ductility of the brazed joint [7,8]. Generally, the melting points of the Ti–Cu–Ni–Zr amorphous filler metals are approximately 100 °C lower than that of the conventional titanium-based filler metals. Therefore, using these filler metals makes it possible to braze the Ti–6Al–4V alloy at a temperature below its β -transus, avoiding the formation of coarse grains and Widmanstätten structure in the brazed joint [9–11].

This study concentrates on an approach to joint C103 and Ti–6Al–4V alloys by vacuum furnace brazing using Ti–20Cu–20Ni–20Zr (wt.%) as the brazing filler metal. The microstructural evolution and shear strength of the brazed joint have been comprehensively studied to clarify the relations between the microstructure and the joint performance.

2. Experimental procedures

Sheets of C103 and Ti–6Al–4V alloys of 2 mm in thickness were used as base metal in this study. Both alloy sheets were cut into $10 \times 10 \times 2$ mm and $15 \times 12 \times 2$ mm by wire-cutting for microstructure analysis and shear-strength test, respectively. These specimens were then polished with wet SiC paper of 100 grit, and then subsequently cleaned using an ultrasonic bath with acetone as the fluid before brazing. A Ti–20Cu–20Ni–20Zr (wt%) brazing alloy was used as the brazing filler metal, in the form of foil with the thickness of 60 μ m and 80 mm wide. This brazing filler metal has solidus and liquidus temperature of 848 °C and 856 °C, respectively. The brazing foil was cleaned by acetone before brazing. Furnace brazing was conducted in a vacuum of 1×10^{-6} Torr at a temperature range between 950 °C and 1000 °C for 5–20 min. The heating rate was set at 20 °C/min throughout the study. All specimens were preheated to 830 °C and holded at this temperature for 3 min before being heated up to the brazing temperature. After brazing, all brazed specimens were furnace-cooled from the brazing temperature to room temperature. Following bonding, specimens were cut by a low speed diamond saw, and then were processed by a series of standard metallographic procedures before the microstructure investigation. The cross-sections of the brazed specimens were investigated using a Phillips XL-30 scanning electron microscope (SEM) with an accelerating voltage of 20 kV. The quantitative chemical composition analysis of the

brazed joints was conducted using a JEOL JXA-8800M electron probe microanalyzer (EPMA) equipped with a wavelength dispersive spectrometer (WDS). The spot size of the EPMA was 1 μ m and its operation voltage was 20 kV. The shear test was conducted with a Shimadzu AG-10 universal testing machine at a constant speed of 1 mm/min. The shear test specimens had a width of 15 mm with an overlap length of 2 mm and a thickness of 2 mm. The phase identification of the brazed joint was carried out using a Philips PW 1710 X-ray diffractometer with Cu-K α radiation from 10° to 90°. To identify the phases in the brazed joint, the specimen brazed at 1000 °C for 5 min was cut parallel with the joint interface to remove most C103 base metal (please see Fig. 1). The residual C103 base metal on the surface of the specimen was polished off by SiC paper of 2400 grit, so that the cross-section of region D could be appeared on the surface of the specimen and was identified by X-ray diffractometer. Following the X-ray analysis of region D, the specimen was polished gently to obtain the cross-section of the next region (region C). The SEM observation was conducted to confirm those regions appeared on the surface of the specimen. X-ray analysis of each region was conducted sequentially until the four regions were all identified.

3. Results and discussion

Fig. 1 shows a back-scattered electron image (BEI) of the C103/TiCuNiZr/Ti–6Al–4V joints brazed at 1000 °C for 5 min. Additionally, EPMA chemical analyses of four regions in the brazed joint are also included. This photograph presents the microstructural transition from the substrates throughout the brazed joint. These characteristic regions are formed by atomic diffusion, including isothermal solidification and solid-state diffusion during brazing [12]. Based on the microstructural morphology and the chemical composition, all microstructures in the joint interface can be classified into four characteristic regions. The formation of these regions indicate that the strong reaction of the molten filler metal with the Ti–6Al–4V and C103 base metals during brazing. Fig. 2 displays the XRD results of the four regions in the cross-sectioned joint interface brazed at 1000 °C for 5 min. According to the XRD and EPMA results, four regions are discussed as follows:

(1) Region A is an acicular Widmanstätten structure. During brazing, diffusion of Cu and Ni atoms from the liquid filler metal into the original equiaxed α plus intergranular β base metal transformed the structure into the β phase because the Cu and Ni elements are the β -Ti stabilized for the Ti–6Al–4V alloy. On cooling from β phase at high temperature, acicular α phase nucleates at the grain boundary of β phase and grows along $[110]_{\beta}$ directions, leaving untransformed β phase [13,14].

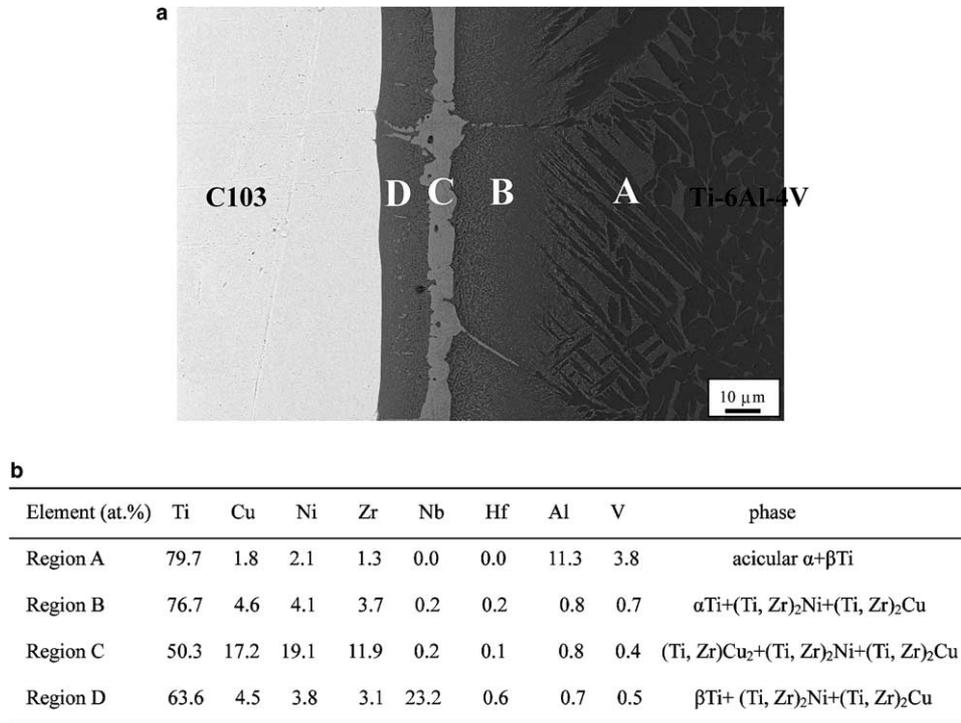


Fig. 1. (a) The microstructure of the C103/TiCuNiZr/Ti-6Al-4V joint brazed at 1000 °C for 5 min, showing four characteristic regions of the joint interface. (b) The EPMA analyses of the four regions.

(2) Region B is a eutectoid microstructure containing Ti and a small amount of Ni, Cu and Zr at the Ti-6Al-4V side in the brazed joint. The XRD results indicate that region B is comprised of α -Ti and Ti₂Ni and Ti₂Cu. However, because Ti₂Ni and Ti₂Cu have some solubility of Zr [7,8,16], these phases should be (Ti,Zr)₂Ni and (Ti,Zr)₂Cu, in fact. These phases should be the results of the reaction of dissolving elements of base metal with Cu, Ni and Zr elements of the liquid filler metal. Fig. 3(a) shows the very fine lamellar eutectoidal microstructure of region B, consisting of dark lamellate α -Ti and bright lamellate (Ti, Zr)₂Ni or (Ti, Zr)₂Cu.

(3) Region C is a central brazed layer between regions B and D. The composition of region C is almost identical with that of the original filler metal. The XRD results show that region C consists of a large amount of inter-metallic compounds, such as (Ti, Zr)Cu₂, (Ti, Zr)₂Ni and (Ti, Zr)₂Cu.

(4) Region D is a eutectoid microstructure containing Nb and Ti and a small amount of Ni, Cu and Zr at the C103 side in the brazed joint. In this region, the concentration of the β -Ti stabilizer, such as Nb, Cu and Ni (Nb especially), is high enough to suppress α -Ti formation during cooling. Therefore, this region is comprised of β -Ti and (Ti, Zr)₂Ni and (Ti, Zr)₂Cu, as shown in Fig. 2(d). The formation reason of this region D is similar to that of the region B. Fig. 3(b) shows the very fine lamellar eutectoidal structure of this region, consisting of dark lamellate β -Ti and bright lamellate (Ti, Zr)₂Ni or (Ti, Zr)₂Cu.

Fig. 4(a)–(d) shows the SEM back-scattered electron images (BEI) of the C103/TiCuNiZr/Ti-6Al-4V joints brazed at 950 °C for 5, 10, 15 and 20 min, respectively. The thickness of both sides of eutectoid regions increases with the increasing brazing time, whereas the width of the central brazed layer reduces with the increasing brazing time. As the brazing time increases to 20 min at 950 °C, the central brazed layer disappears and only the eutectoid regions can be observed in the brazed joint. Additionally, few acicular Widmanstätten structure form in front of Ti-6Al-4V base metal, because the brazing temperature is lower than the β -transus temperature of Ti-6Al-4V.

Fig. 5(a)–(d) shows the SEM back-scattered electron images (BEI) of the C103/TiCuNiZr/Ti-6Al-4V joints brazed at 1000 °C for 5, 10, 15 and 20 min, respectively. It is expected that the higher brazing temperature and/or longer brazing time would enhance metallurgical reaction resulting from the increased dissolution and diffusion. Therefore, the central brazed layer in the joint disappears within 10 min as the brazing temperature is raised to 1000 °C. Moreover, the microstructural observations reveal that increasing the brazing time at 1000 °C would promote the formation of coarse acicular Widmanstätten structure and the coarse grains of Ti-6Al-4V.

Fig. 6 shows the shear strength of the C103/TiCuNiZr/Ti-6Al-4V joints plotted against the brazing time at different brazing temperatures. The testing results reveal that the shear strength of the specimen brazed at

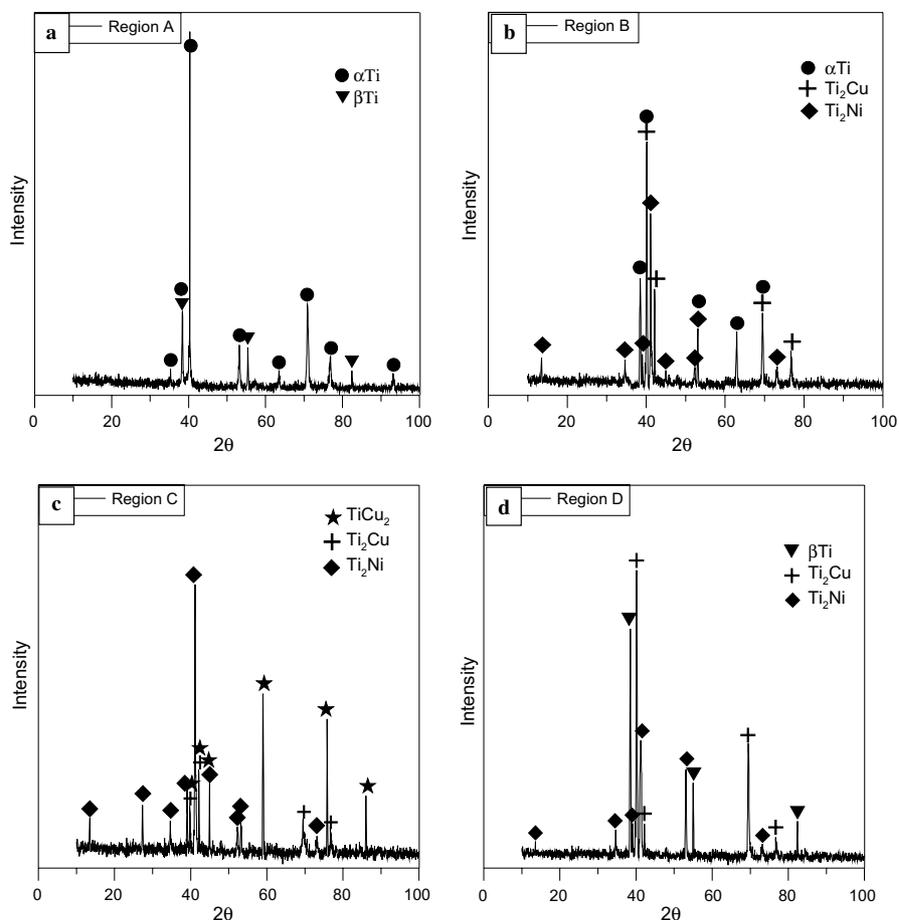


Fig. 2. The X-ray analysis of the four regions in the cross-sectioned joint interface brazed at 1000 °C for 5 min: (a) the Widmanstätten structure (region A), (b) the eutectoidal structure (region B), (c) the central brazed layer (region C), (d) the eutectoidal structure (region D).

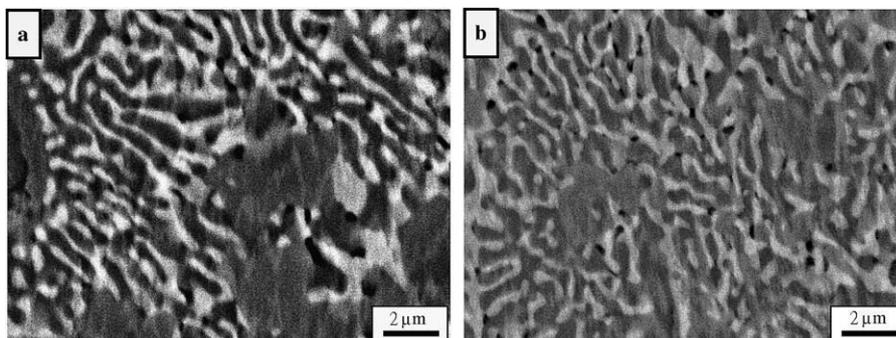


Fig. 3. The fine lamellar eutectoid structure of (a) region B and (b) region D in the C103/TiCuNiZr/Ti–6Al–4V joint brazed at 1000 °C for 5 min.

950 °C increases with the increasing brazing time, and the joints were fractured in the base metal for the brazing time of 20 min. However, as the brazing time is shorter than 15 min, the central brazed layer would remain in the brazed joint, and the brazed joints with low strengths less than 270 MPa were fractured in central brazed layer and the feature of the fracture surface was quite brittle, as shown in Fig. 7. It was found from these results that the central brazed layer consisting of

brittle intermetallic compounds, such as $(\text{Ti}, \text{Zr})\text{Cu}_2$, $(\text{Ti}, \text{Zr})_2\text{Ni}$ and $(\text{Ti}, \text{Zr})_2\text{Cu}$, caused low shear strengths of the joints. Therefore, to obtain the sufficient shear strengths for the C103/TiCuNiZr/Ti–6Al–4V joints, disappearance of the central brazed layer consisting of brittle intermetallic compounds is needed.

Additionally, the evolution of shear strength of the specimen brazed at 1000 °C differs markedly from that of the specimens brazed at 950 °C. The shear strength of

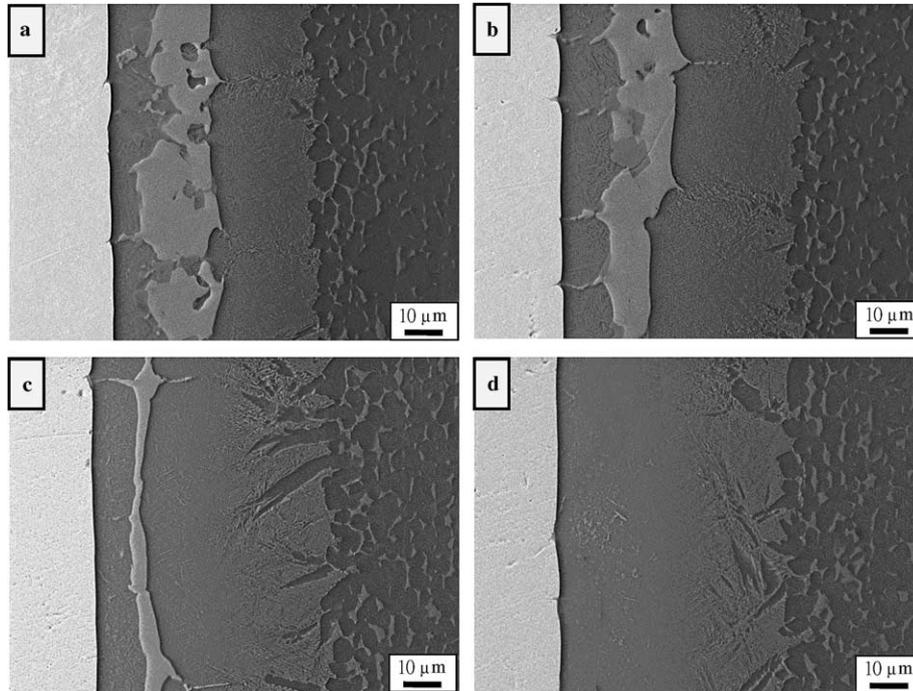


Fig. 4. The back-scattering electron images (BEIs) of C103/TiCuNiZr/Ti-6Al-4V specimens brazed at 950 °C for (a) 5, (b) 10, (c) 15, and (d) 20 min.

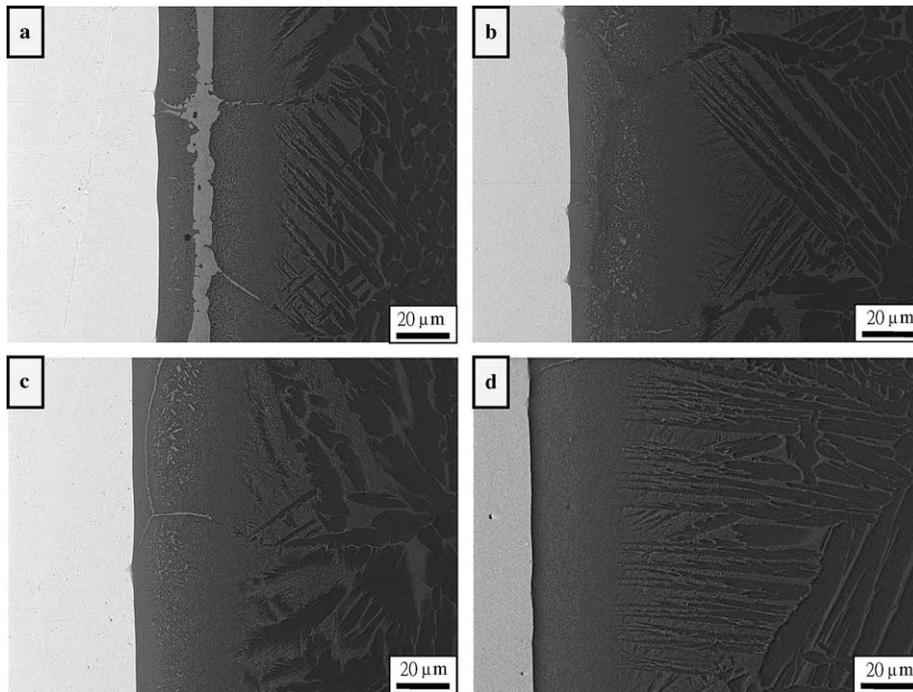


Fig. 5. The back-scattering electron images (BEIs) of C103/TiCuNiZr/Ti-6Al-4V specimens brazed at 1000 °C for (a) 5, (b) 10, (c) 15 and (d) 20 min.

joint reduces with the increasing brazing time as the brazing temperature raises to 1000 °C. The microstructural observations, as shown in Fig. 5, reveal that no central brazed layer exists in the joint interface of specimens brazed at 1000 °C as the brazing time prolongs to 10 min. However, a large amount of coarse acicular Wid-

manstätten structure formed in the joint, because the brazing temperature exceeds the β -transus temperature of Ti-6Al-4V. Further prolonging the brazing time makes the Widmanstätten structure become longer and coarser, decreasing the shear strength of joints [8,10]. This is the reason why the shear strength of the specimen

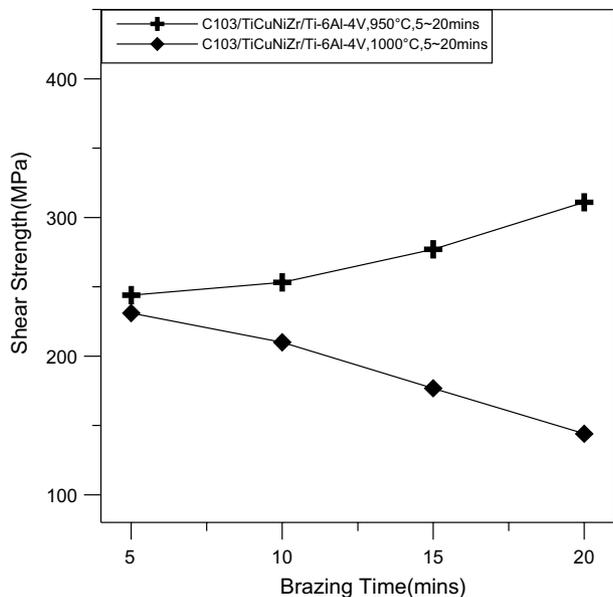


Fig. 6. The shear strength of C103/TiCuNiZr/Ti-6Al-4V joint brazed under various brazing conditions.

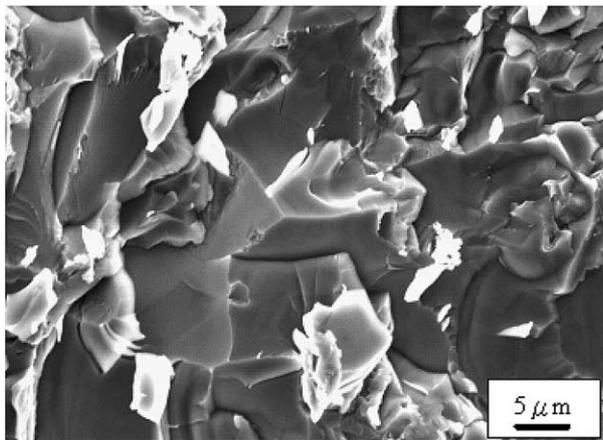


Fig. 7. Fractography of C103/TiCuNiZr/Ti-6Al-4V joint brazed at 950 °C for 5 min after shear-strength test.

brazed at 1000 °C reduces with the increasing brazing time, as shown in Fig. 6.

4. Conclusion

Microstructural evolution and shear strength of the brazed Ti-6Al-4V and C103 joint using the Ti-20Cu-20Ni-20Zr (wt.%) filler metal has been studied in this research. The primary conclusions are summarized as below:

1. Based on the microstructural observation, the central brazed layer consisting of intermetallic compounds,

(Ti,Zr)Cu₂, (Ti,Zr)₂Ni and (Ti,Zr)₂Cu, formed at the brazing temperature of 950 °C and remained for brazing time of 15 min. As the brazing time extended to 20 min, the central brazed layer disappeared and only the eutectoidal region was observed. As the brazing temperature increased to 1000 °C, the central brazed layer disappeared within the brazing time of 10 min and only the eutectoid and Widmanstätten regions could be found in the joint interface.

2. The shear strength of joint brazed at 950 °C increases with the increasing brazing time because of the disappearance of the brittle central brazed layer. The brazed joints with low shear strength less than 270 MPa were fractured in the central brazed layer because of the brittleness of the intermetallic compounds existed in the central brazed layer. The shear strength of the joint brazed at 1000 °C decreases with the increasing brazing time, because of the formation of coarse acicular Widmanstätten structure in the joint.
3. To obtain a high shear strength of joint, both the central brazed layer and the coarse Widmanstätten structure should be avoided forming in the interface of brazed joint.

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