Materials Science Forum Vols. 730-732 (2013) pp 835-840
Online available since 2012/Nov/12 at www.scientific.net

doi:10.4028/www.scientific.net/MSF.730-732.835

Microstructure and Strength of γ-TiAl alloy / Inconel 718 brazed joints

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Keywords: γ-TiAl based alloy; Ni superalloy; Brazing; Mechanical Properties.

Abstract. Intermetallics and superalloys brazing development is a current topic owing the extending use of these alloys in industrial applications. In this work a γ-TiAl alloy was joined to Inconel 718 by active metal brazing, using Incusil-ABA as filler. Joining was performed at 730 °C, 830 °C and 930 °C, with a 10 min dwelling time. The interfaces were characterized by Scanning Electron Microscopy (SEM), Energy Dispersive X-ray Spectroscopy (EDS) and Electron Backscatter Diffraction (EBSD). For all processing conditions, the reaction between the base materials and the braze alloy produced multilayered interfaces. For all processing temperatures tested (Ag), (Cu), $AlNi_2Ti$ and $AlCu_2Ti$ were identified at the interface. Raising the brazing temperature increased the thickness of the interface and coarsened its microstructure. The increase of the extension of the interface was essentially due to the growth of the reaction layers formed near each base material, which were found to be mainly composed of intermetallic compounds. The mechanical behavior of the joints, at room temperature, was assessed by microhardness and shear tests. For all processing conditions the hardness decreases from periphery towards the Ag-rich centre of the joints. Brazing at 730 °C for 10 min produced the joints with the highest average shear strength (228±83 MPa). SEM and EDS analysis of the fracture surfaces revealed that fracture of joints always occurred across the interface, preferentially through the hard layer, essentially composed of $AlNi_2Ti$, resulting from the reaction between Inconel 718 and the braze alloy.

Introduction

Intermetallic alloys are an emerging class of materials, which are particularly appealing for applications where weight reduction and enhanced temperature capability are critical inputs concerning materials selection [1-5]. Alloys based on the γ-TiAl intermetallic compound are promising candidates for replacing titanium alloys and nickel superalloys for structural applications in the temperature range of 400-800 ºC, because of their attractive properties like high-temperature strength retention, low density (3.7–3.9 $g.cm^{-3}$), excellent burn resistance, high modulus of elasticity and good oxidation resistance [4,5].

Since nearly all potential application of γ-TiAl alloys will require joining, the research and development of adequate techniques to join these alloys to other materials or to themselves is crucial to integrate them into functional structures as well as to expand their market [1,3,4,6-8]. Bonding of dissimilar materials is the most complex situation, since mismatches between base materials thermal and mechanical properties usually originate residual stresses that may induce nucleation and propagation of cracks at the interfacial region of the joints. To ensure joint integrity, these residual stresses must be accommodated adequately by the interfacial region [7,9], through the formation of reaction products with suitable chemical composition, morphology and distribution across the interface. Diffusion bonding, diffusion brazing and brazing have been reported as suitable methods to produce TiAl alloys joints [1,7,8]. Among these techniques, brazing is the most straightforward joining procedure; it presents several advantages, namely: the direct mixture of the base materials can be prevented, the residual stresses may be efficiently accommodated by the interface if ductile brazing fillers are used, joining may be performed at relatively low temperatures and without the need of applying elevated pressures to promote bonding [7,8,10]. Another advantage of brazing is the possibility of selecting the interlayer, according to the characteristics of the base materials, among a wide variety of commercially available compositions [11]. Successful joining will depend of the interface characteristics, namely of its mechanical behavior, which in turn is highly dependent of the microstructure developed at the interface, and on how the discontinuity of properties is accommodated. For a given system, the microstructure is determined by the joining processing variables, such as the brazing temperature, the heating and cooling rates, and the brazing holding stage [2-4,10-12]. Therefore, assessing the influence of these variables on the chemical, microstructural and mechanical features of the interfaces is crucial to optimize the joining procedure.

The present study focuses on the influence of the processing temperature upon the chemical and microstructural features of the interfaces and the mechanical properties of γ-TiAl/Inconel 718 joints, brazed by active metal brazing, using Incusil-ABA as filler alloy.

A successful industrial application of γ-TiAl alloys consists in the production of turbocharger rotors for passenger vehicles. Brazing the γ-TiAl rotor to a Ni-based superalloy insert, using a Ag-Cu eutectic filler, is reported to be used in the manufacturing process of this high performance automotive engine component [6]. In comparison to the Ag-Cu eutectic filler, Incusil-ABA allows to produce joints at lower temperatures, since its liquidus temperature (715 ºC) is 65 ºC lower. Therefore, joining using Incusil-ABA may be achieved faster (given the same brazing dwell stage, heating and cooling rates) and with less consumption of energy.

Materials and Experimental Procedures

Samples of γ-TiAl alloy (Ti47Al2Cr2Nb, at%) with 10x10x10 mm and of Inconel 718 (53Ni19Cr19Fe, wt%) and with a diameter of 10 mm and 12 mm height, were cut with a diamond saw and then polished with SiC emery paper down to 1200 mesh. A 100 μm Incusil-ABA (Ag27.5Cu12.5In1.25Ti, wt.%) foil was used as filler. Prior to joining, all materials were degreased in acetone with ultrasonic agitation, dried in air and then assembled into a sandwich type. Brazing was performed in a high-vacuum furnace at 730, 830 and 930 °C, with a dwelling time of 10 min. The heating and the cooling rates were fixed at 5 $ºC.min^{-1}$. The vacuum level was better than 10^{-4} mbar during the brazing thermal cycle. A pressure of 50 Pa was applied to the joining assemblage, by means of a stainless steel mass, in order to promote an intimate contact between all the adjoining surfaces of samples. In order to perform the microstructural and chemical characterization of the interfaces, cross-sections of the joints were prepared using standard metallographic techniques. The interfaces were characterized by SEM, EDS and EBSD analysis. EBSD patterns were used to assess the crystallographic features of some of the phases detected at the interfaces. Mechanical properties at room temperature were evaluated by shear and microhardeness tests. Microhardeness tests were performed with a Vickers indenter, using a load of 100 mN during 15 seconds. The shear strength of the joints was assessed by testing four samples for each brazing condition. Additional details on the shear test apparatus are given elsewhere [13]. The fracture surfaces of joints were analyzed by SEM/EDS.

Microstructural and Chemical Characterization

All processing temperatures used in this investigation led to the formation of multilayered interfaces, as it can be observed in Fig. 1. The interface can be divided into three characteristics reaction layers. For convenience these layers will be identified by letters (A, B and C). Layers A and C result from the reaction between the filler metal and Inconel and γ-TiAl alloy, respectively; layer B consists essentially of the residual filler alloy.

Raising the brazing temperature from 730 to 930 ºC increased the extension of the interface from about 65 to 100 μm, respectively and coarsened the microstructure. For higher brazing temperatures, the thickness of the interface increases mainly due to the growth of the reaction layer formed near each base material. The EDS composition maps presented in Fig. 2 show that in the course of joining intense diffusion occurred across the interface. Ag is manly detected at the central zone of the interface and in the vicinity of Inconel 718. In spite of not being presented in the figure, In presents a distribution map similar to that of Ag. The composition maps also show that both Ti and Al, who diffused only from the γ-TiAl alloy, mainly concentrate in layers formed adjacently to each base material. Cu is detected throughout the interface but presents higher contents near the γ-TiAl alloy. Ni is essentially detected in regions located near Inconel 718.

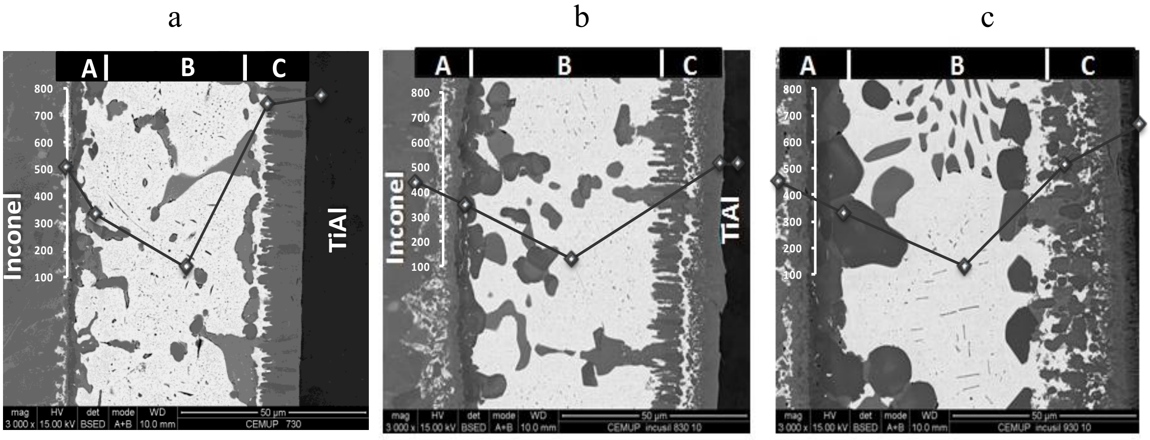

Fig. 1 BSE images of the interfaces and hardness profile resulting from brazing at: a) 730 ºC, b) 830 ºC and c) 930 ºC

The central zone of the interface (layer B) consists of a Ag-rich bright matrix (78Ag17In5Cu, at.%) in which coarse dark particles are dispersed that are rich either in Cu (94Cu3In2Ni1Ag, at.%) or in Cu, Ti and Al (54Cu25Ti19Al2In, at.%). Diffusion of Ag towards Inconel induced in layer A the formation of small Ag-rich particles with a complex chemical composition (53Ag15Fe10In6Cr, at.%) alloyed with Ti, Cu, Ni and Mo; BSE images suggest that grain boundary diffusion was the predominant path for Ag diffusion. This layer is also composed of a continuous dark zone (23Ni24Ti21Al32Cu, at.%) formed contiguously to layer B. Layer C, for lower brazing temperatures, presents a columnar like morphology, which can be clearly observed at the zones neighbouring layer B rather than near the γ-TiAl alloy, is rich in Cu, Al and Ti (54Cu24Ti22Al, at.%). Bright Ag-rich zones, with a chemical composition similar to that of the matrix of layer B, were also observed in this layer.

EDS does not allow the identification of the phases that constitute the interface. However, equilibrium phase diagrams could be used in conjunction with SEM and EDS as predictive tools of the nature of the phases formed at the interface. Plotting the composition of the dark zones of layer C, which are is essentially composed of Al, Cu and Ti, on the Al-Cu-Ti equilibrium ternary phase diagram [14], indicates the possible formation of $AlCu_2Ti$. Based on the EDS composition and on the Ag-Cu-In ternary diagram [14], the bright matrix observed in layer B should consist of an Ag solid solution, while the coarse dark particles in this layer should be a Cu solid solution. The continuous dark zone in layer A is essentially composed by Al, Cu, Ti and Ni. Since the equilibrium quaternary phase diagram Al-Cu-Ni-Ti has not yet been established, the Al-Ni-Ti and Al-Cu-Ti equilibrium ternary phase diagrams [14] can be used to bypass this problem, as long as one adds Cu and Ni contents. This approach, which is analogous to the one of Lee *et al.* [15,16], is based on the fact that Ni and Cu atoms behave in a similar way, since both have similar atomic sizes and electronegativities, the same crystalline structure and unlimited solid solubility, enables to suggest the formation $Al(Cu,Ni)_2Ti$.

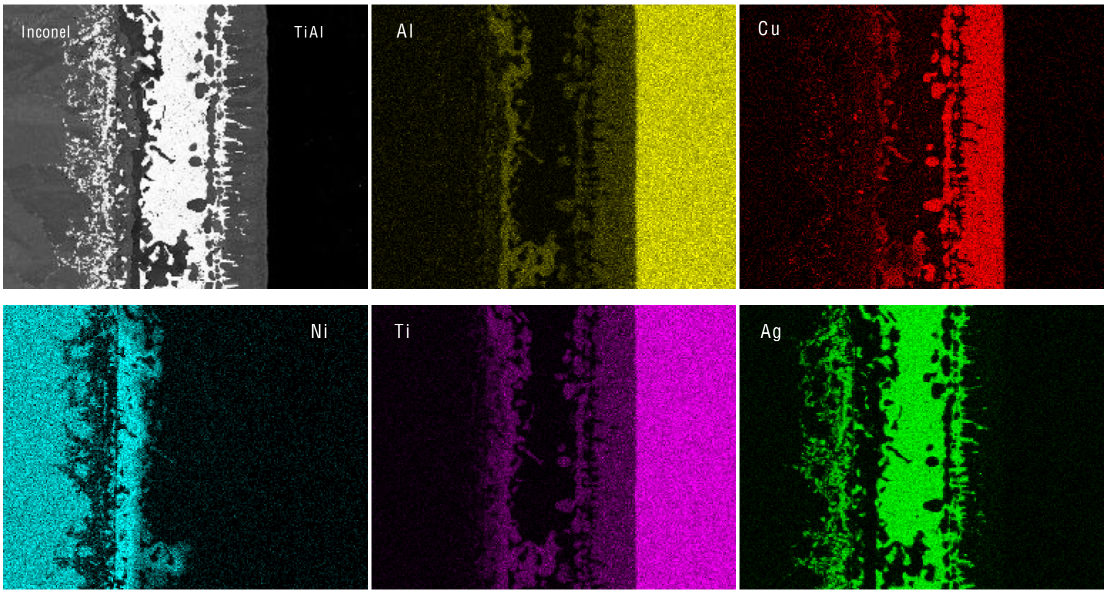

Fig. 2 BSE image and EDS mapping of the interface resulting from brazing at 830 ºC

EBSD analysis was used to assess the identification of some of the phases detected at the interfaces. Fig. 3 shows the result of an EBSD analysis performed on the region marked by the dot on the BSE image of the interface resulting from brazing at 730 ºC (Fig. 3a); the EBSD pattern (Fig. 3b), the indexation of the Kikuchi lines (Fig. 3c) and the EDS analysis (Fig. 3d) are also presented. The Kikuchi lines of the analysed region are indexed as characteristics of $AlCu_2Ti$. Similar EBSD analysis, performed on the dark zones of layers A and C, identified $AlNi_2Ti$ and $AlCu_2Ti$ as phases formed in the reaction layers located near to Inconel 718 and the γ-TiAl alloy, respectively. Further, EDS and EBSD analysis indicated that the nature of the products formed at the interface is independent of the temperature at which brazing was carried out in this investigation.

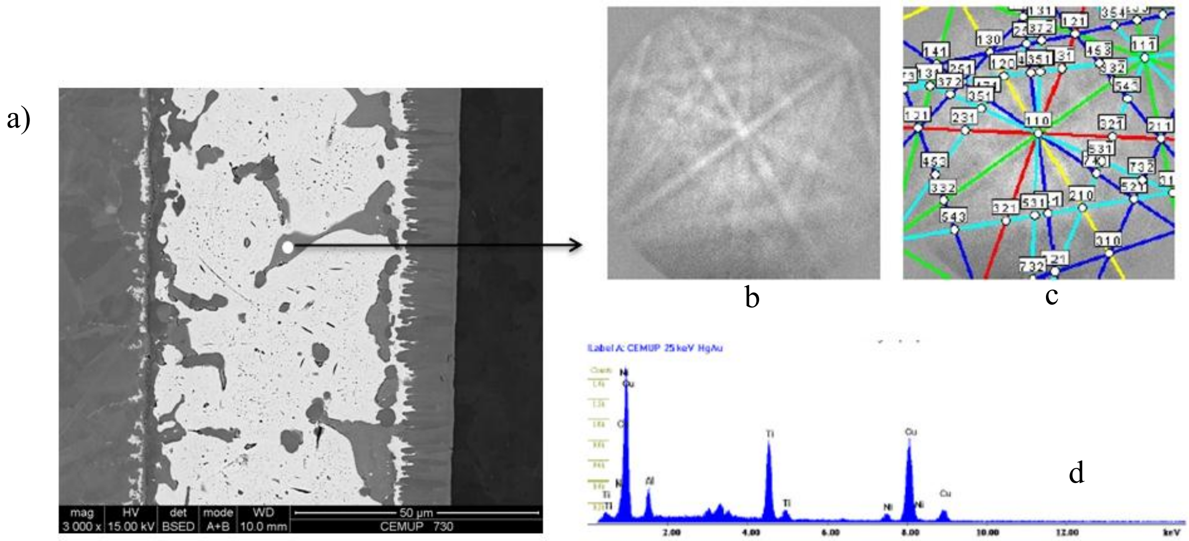

Fig. 3 EBSD/EDS/SEM analysis of the interface resulting from brazing at 730 ºC. a) BSE image, b) EBSD pattern, c) indexing of Kikuchi lines that identify $AlCu_2Ti$ phase, and d) EDS analysis

Mechanical properties

The hardness profile across the interface for each brazing temperature is presented in Fig.1. Each point represents the average of at least three indentations. The hardness decreases from the periphery towards the centre of the joints. For instance, after brazing at 730 ºC layer B (134±16 HV0.1) is considerably softer than both base materials, while the layers A and C (738±39 HV0.1 and 771±171 HV0.1, respectively) are harder.

The increase of the brazing temperature from 730 to 830 ºC results in the decrease of the hardness of layers A and C. As it can be observed in Fig.1, the hardness of layers A and C is 512±97 HV0.1 and 512±110 HV0.1, respectively (hardness values close to the typical hardness of Inconel 718), whilst the hardness of the Ag-rich region remained the same. The hardness profile of 930 ºC brazed joint is similar to the one measured for brazing at 830 °C. However, the thickness of the hardest layers increases, whereas that of the softer (layer B) decreases.
The average shear strength of joints is 228±83, 187±55 and 164±76 MPa, for brazing at 730, 830 and 930 ºC, respectively. A large dispersion of results is observed for all brazing temperatures; for instance, the strength of joints brazed at 730 ºC varies between 132 and 280 MPa.
In order to understand the shear response of the brazed joints, top and cross-sections of the fracture surfaces were observed by SEM and analyzed by EDS. For all processing conditions, joints always fractured across the interface, preferentially through layer A. Fig. 4 shows the cross-section of the fracture surface of a high strength joint (268 MPa). As it can be observed in the figure, the joint fractured mainly across the reaction layer located near Inconel 718. At higher magnification (see Fig. 4b and 4c) it is clear that facture occurred through both the Ag-rich and the $AlNi_2Ti$ intermetallic phases formed in layer A.
Observing the fracture surface with more detail, allows the identification of different mechanisms of fracture. Fracture surfaces of high strength joints present characteristic features of both brittle and ductile fracture: cleavage facets (Fig. 5a), which are typical features of a brittle fracture, when fracture occurs through intermetallic particles formed in layer A; and dimples (Fig. 5b) that indicate a ductile behaviour, when fracture occurs through the Ag-rich regions located near Inconel 718. Microvoids, similar to shrinkage defects, are often observed in low strength joints (Fig. 5c).

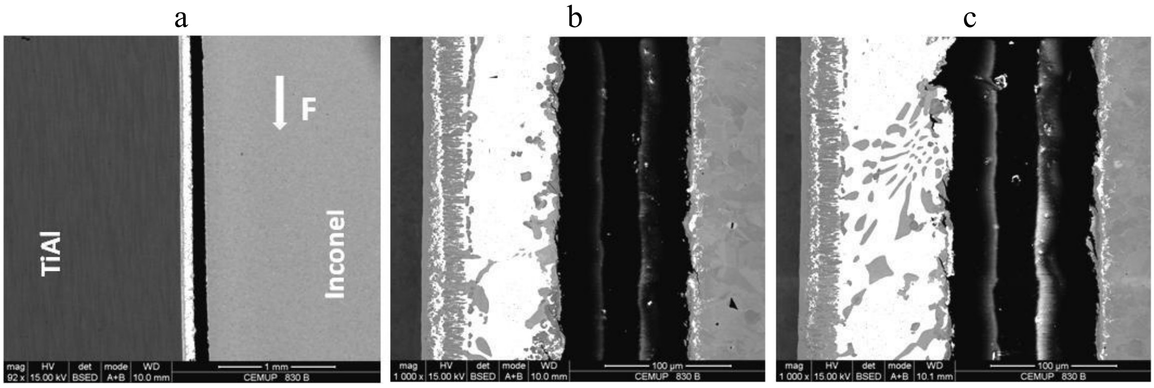

Fig. 4 BSE images of cross-sections of fracture surfaces of a high strength joint brazed at 830 ºC: a) low magnification, b) and c) high magnification

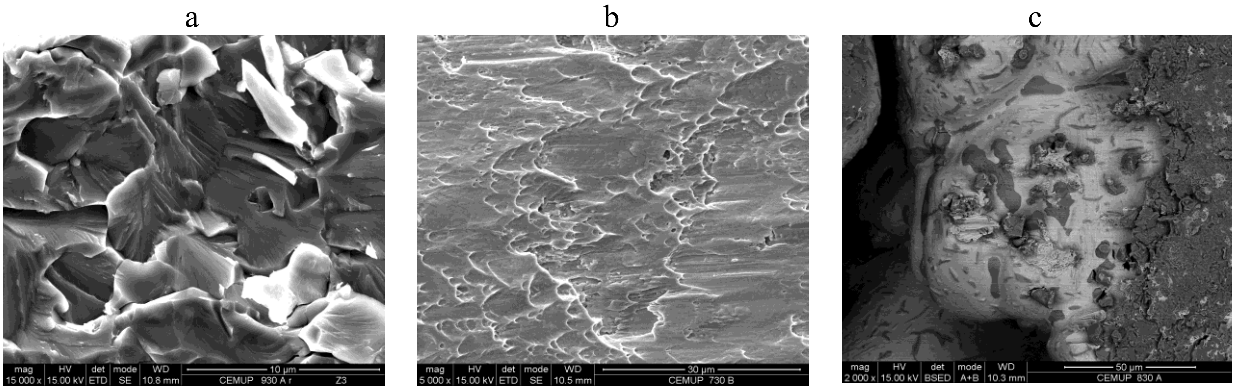

Fig. 5 Fracture surfaces high strength and low strength joints: a) SE image of the fracture surface of a high strength joint brazed at 930 ºC, where typical cleavage fracture is observed. b) SE image of the fracture surface of a high strength joint brazed at 730 ºC, showing a ductile fracture feature. c) BSE image of the fracture surface of a low strength joint brazed at 830 ºC, showing microshrinkages

Conclusions

The joining by active metal brazing of γ-TiAl alloy to Inconel 718, at 730, 830 and 930 °C, with a dwell time of 10 minutes, using Incusil-ABA as filler was investigated and the following conclusions can be drawn:

- For all brazing temperatures tested multilayered interfaces, apparently free of cracks, were produced;
- Raising the brazing temperature increased the extension of interface and coarsens its microstructure;
- The main phases detected in each reaction layer were as follows, starting from the Inconel 718 side of the interface towards the γ-TiAl alloy side: (Ag) + $AlNi_2Ti$ (layer A) / (Ag) + (Cu) + $AlCu_2Ti$ (layer B) / $AlCu_2Ti$ + (Ag) (layer C);
- Exception made to the central zone, the interface is harder than both base materials. However, the hardness of the reaction layers formed contiguously to the base materials tends to decrease as the brazing temperature is raised, but their thickness increases;
- The maximum average shear strength (228±83 MPa) was displayed by joints brazed at 730 ºC. Increasing the processing temperature decreased the shear strength of the joints;
- Joints always fractured through the interface, preferentially through the reaction layer formed adjacently to Inconel 718;
- Fracture surfaces exhibit regions with characteristics features of ductile and brittle fracture. Microshrinkage defects are observed on fracture surfaces of low strength joints;
- Further improvement is needed to reduce the dispersion of the joints shear strength values in order to increase the reliability of the joining procedure.

Acknowledgments This work was supported by “FCT-Fundação para a Ciência e Tecnologia” through the project PTDC-CTM-69646-2006.

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Advanced Materials Forum VI
10.4028/www.scientific.net/MSF.730-732

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