

Joining Ti-47Al-2Cr-2Nb with a Ti-Ni braze alloy

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Keywords: Titanium Aluminides, Joining, Diffusion Brazing, Interfaces.

Abstract. Ti-47Al-2Cr-2Nb was joined by diffusion brazing using a Ti-Ni clad-laminated filler alloy. Brazing was performed in the temperature range of 1000 to 1200°C for 10 min. The microstructure and the chemical composition of the interfaces were studied by scanning electron microscopy (SEM) and by energy dispersive X-ray spectroscopy (EDS), respectively. The reaction between the γ -TiAl alloy and the liquid produced interfaces essentially composed of α_2 -Ti₃Al and TiNiAl intermetallics. When joints are brazed at 1150 and 1200°C, the γ -TiAl phase is also detected at the interface.

Introduction

Intermetallic alloys are an emerging class of materials that may have a wide variety of applications, especially when weight reduction and enhanced temperature capability are critical inputs for structural designs. Among these alloys, two phase γ -based TiAl alloys are advancing towards implementation on structural components of aircraft, space and automobile engines [1-4]. The use of γ -TiAl alloys as structural materials, will require the development of reliable and cost-effective joining techniques. Diffusion brazing [5] and diffusion bonding [6-8] seem to be the most straightforward techniques to produce joints with mechanical properties comparable to those of γ -TiAl alloys at high temperature. Diffusion bonding causes less disruption of the surface of the components to be joined and more homogeneous interfaces than does diffusion brazing. Ideally, diffusion bonded joints have the same composition and structure of the parent material, and therefore the same mechanical properties. However, diffusion brazing offers some advantages over diffusion bonding: joining is carried out at lower temperature without the need of applying elevated pressures. The mechanical behaviour diffusion brazed joints is highly dependent on the features of the interface, namely on the chemical and mechanical characteristics of the reaction products and their distribution across the interface; these are in turn affected by the processing variables such as the joining temperature, the joining holding stage and the heating and cooling rates. Therefore, it is crucial to assess the influence of these variables in order to optimise the joining of γ -TiAl alloys by diffusion brazing.

Materials and Experimental Procedures

The γ -TiAl alloy (Ti-47Al-2Cr-2Nb, at.%) used in this investigation was produced from gas atomised elemental powders by Crucible Research and has a duplex microstructure. Samples of the intermetallic alloy with 10 x 7 x 3 mm were cut with a diamond saw and then ground with SiC emery paper down to 1200 grade. A Tini67 foil (Ti-33Ni, wt.%) with a thickness of 50 μ m was used as filler metal. Tini67 is a clad-laminated alloy that consists of a sandwich of a 12 μ m Ni foil between two 19 μ m Ti foils. Prior to joining both alloys were degreased in acetone with ultrasonic agitation, dried in air and then assembled into a sandwich type. A small pressure (50 Pa) was applied to the γ -TiAl/Tini67/ γ -TiAl assemblage during the brazing thermal cycle. Joining was conducted at 1000, 1100, 1150 and 1200°C for 10 min in a vertical furnace, which was evacuated to a vacuum level better than 10⁻⁴ mbar. The heating and cooling rates were fixed at 3 °C.min⁻¹.

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Joints for microstructural and chemical characterization were cut perpendicularly to the interface and cold mounted in epoxy resin. Final polishing with a solution of 0.04 μm colloidal silica suspension ensured the removal of any deformed layer. The interfaces were examined by SEM and chemically analysed by EDS at an accelerating voltage of 15 keV.

Results and Discussion

The microstructural and chemical analysis of the base material shows that the brazing procedure did not produce noticeable changes on the microstructure of the titanium aluminide alloy and its chemical composition remained unaltered, even in the region close to the joint.

All processing temperatures used in this investigation led to the formation of interfaces composed of two distinct layers, which are labelled as *A* and *B* on the back-scattered electron images (BEI) of the interface presented in Fig. 1. The thickness of the interface increases with the brazing temperature and ranges approximately from 65 to 153 μm , for joining at 1000 and 1200°C, respectively. The sum of the thickness of the two layers *A*, always represents more than 55% of the width of the interface; for joining at 1200°C it nearly represents 75%. Layers *A* and *B* are composed of a mixture of white and dark zones. Since three types of dark zones were detected throughout the interface, they will be designated as D_1 , D_2 and D_3 ; D_1 zones are richer in Cr and Nb and brighter than D_2 zones and these, brighter than D_3 zones (see Fig. 1 b)). D_3 zones are only detected in layer *A* and only for joining at 1150 and 1200°C.

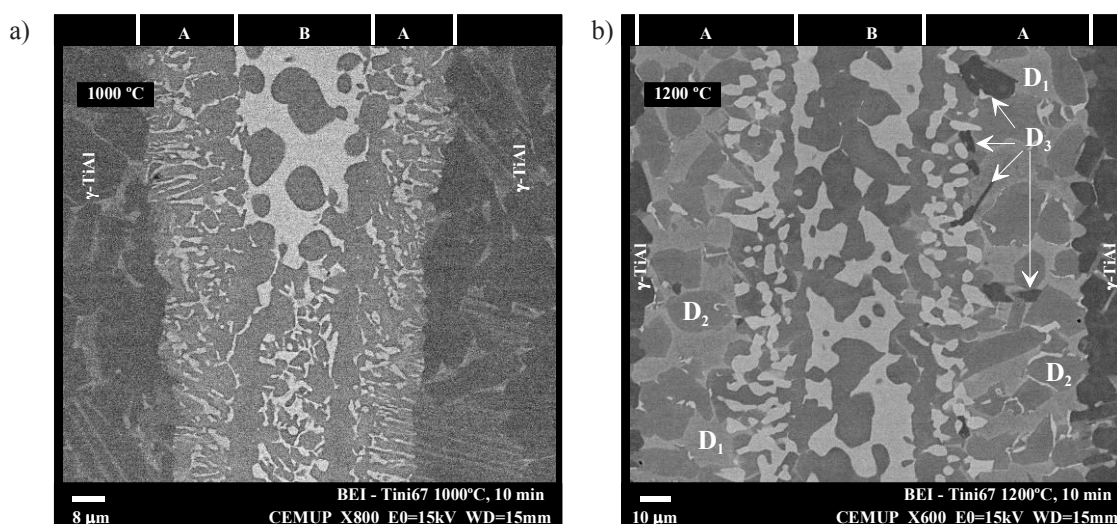


Fig. 1 – BEI of the interface obtained for the two limiting processing temperatures: a) 1000°C; b) 1200°C, showing the three types of dark zones detected at the interfaces.

Independently of the brazing temperature, all white zones are composed of phases containing mainly Ti, Al and Ni and all dark zones only Ti and Al. The composition of the reaction products detected at the interfaces is listed, for the two limiting conditions used in this investigation, in Table 1. Unless stated otherwise, hereafter, all compositions will be expressed in atomic percentage.

In order to evaluate the elemental distribution across the interface, EDS analysis was also performed sequentially from one side of the joint to the other, as it is shown in Fig. 2 a). The elemental distribution across the interface presents a similar profile for all the brazing temperatures. In Fig. 2 b) the profile obtained for joining at 1100°C, as well as the Ti:Al atomic ratio corresponding to each analysed area, are shown. In layer *A*, the Al contents tend to decrease from the periphery towards the centre of the interface, and remain practically constant throughout layer *B*. The distribution profile of Ti presents a similar evolution to that of Al, but it increases abruptly at the transition between layer *A* and layer *B*, where it reaches a sharp maximum. Ni is mainly concentrated at the centre of the interface and its contents tends to decrease from layer *B* to layer *A*. At the transition between these layers, Ni presents a sharp minimum; nearly no Ni is detected at this

zone. Finally, the evolution of the Ti:Al atomic ratio presents almost the same shape as the distribution profile of Ti, except in layer *A* where it increases towards the centre of the joint.

Table 1 – Average chemical composition (at. %) of the reaction products detected at the interfaces for the two limiting joining temperatures (°C) used in this investigation.

Temperature	Layer / Zone	Ti	Al	Ni	Cr	Nb
1000	<i>A</i> / D ₁	57.9	32.2	1.7	5.1	3.1
	<i>A</i> / D ₂	65.7	31.9	0.6	0.9	0.9
	<i>A</i> / White	40.4	36.1	20.9	2.3	0.3
	<i>B</i> / D ₂	73.9	24.4	0.6	0.3	0.8
	<i>B</i> / White	41.5	26.5	29.9	1.7	0.4
1200	<i>A</i> / D ₁	60.5	32.2	1.3	3.7	2.3
	<i>A</i> / D ₂	62.3	34.8	0.4	1.0	1.5
	<i>A</i> / D ₃	53.4	43.5	0.8	0.9	1.4
	<i>A</i> / White	36.5	39.9	19.6	2.6	1.4
	<i>B</i> / D ₂	61.9	35.3	0.5	0.9	1.5
	<i>B</i> / White	36.7	40.7	19.9	1.9	0.8

Since the interfaces are essentially composed of Ti, Al, and Ni, the binary Ti-Al [9] and the ternary Ti-Ni-Al [10] phase diagrams were used in conjunction with the EDS analysis to predict the nature of the reaction products detected at the interfaces. The Ti-Ni-Al isothermal section at 750°C was chosen for this purpose, as it is, within the composition range detected at the interfaces, the isotherm closest to room temperature. In addition, there are no invariant reaction bellows the temperature of the isotherm.

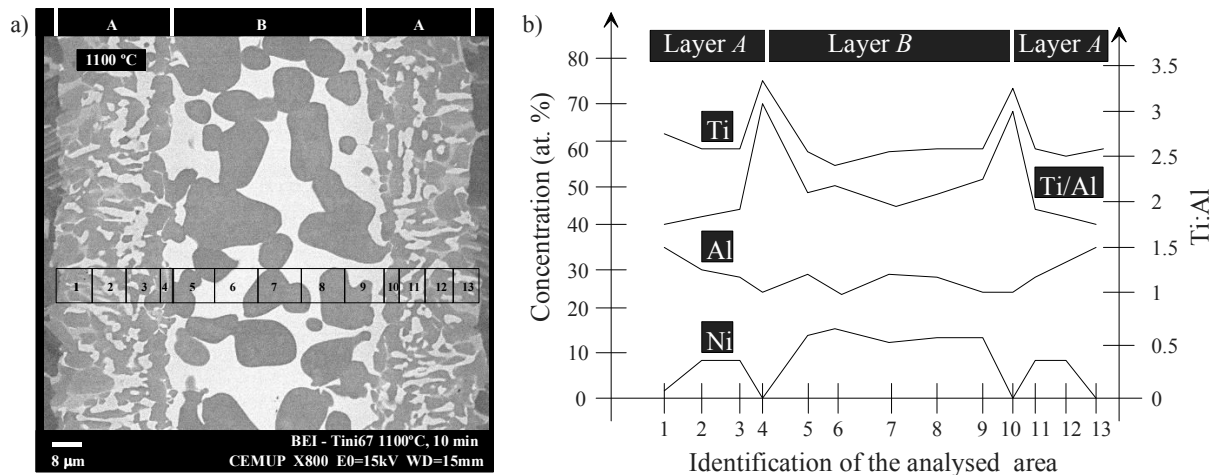


Fig. 2- Distribution profiles of Ti, Al and Ni and evolution of the Ti:Al atomic ratio across the interface: a) BEI of the interface showing the analysed zones used to determine the profiles shown in b).

All dark zones are essentially composed of Ti and Al ($\text{Ti} + \text{Al} > 90\%$), with Ti:Al atomic ratios comprised between 1.8 and 1.9 in D₁ zones, between 1.8 and 3.0 in D₂ zones and being equal to 1.2 in D₃ zones. According to the Ti-Al phase diagram, the $\alpha_2\text{-Ti}_3\text{Al}$ and the $\gamma\text{-TiAl}$ compounds are stable at 500°C for Ti:Al atomic ratios comprised approximately between 1.9 and 4.0 ($\alpha_2\text{-Ti}_3\text{Al}$) and between 0.8 and 1.0 ($\gamma\text{-TiAl}$). Therefore, D₁ and D₂ zones should probably correspond to the α_2 phase and D₃ zones to the γ phase. All white zones are mainly composed of Ti, Ni and Al and seem to be single phase on the BEI of the interfaces. The plots of their composition on the Ti-Ni-Al isotherm, shown in Fig. 3, lay invariably close to the TiNiAl phase field. Therefore, we presume

that all white zones should be essentially composed of TiNiAl (intermetallic phase with structure type MgZn_2 [10]).

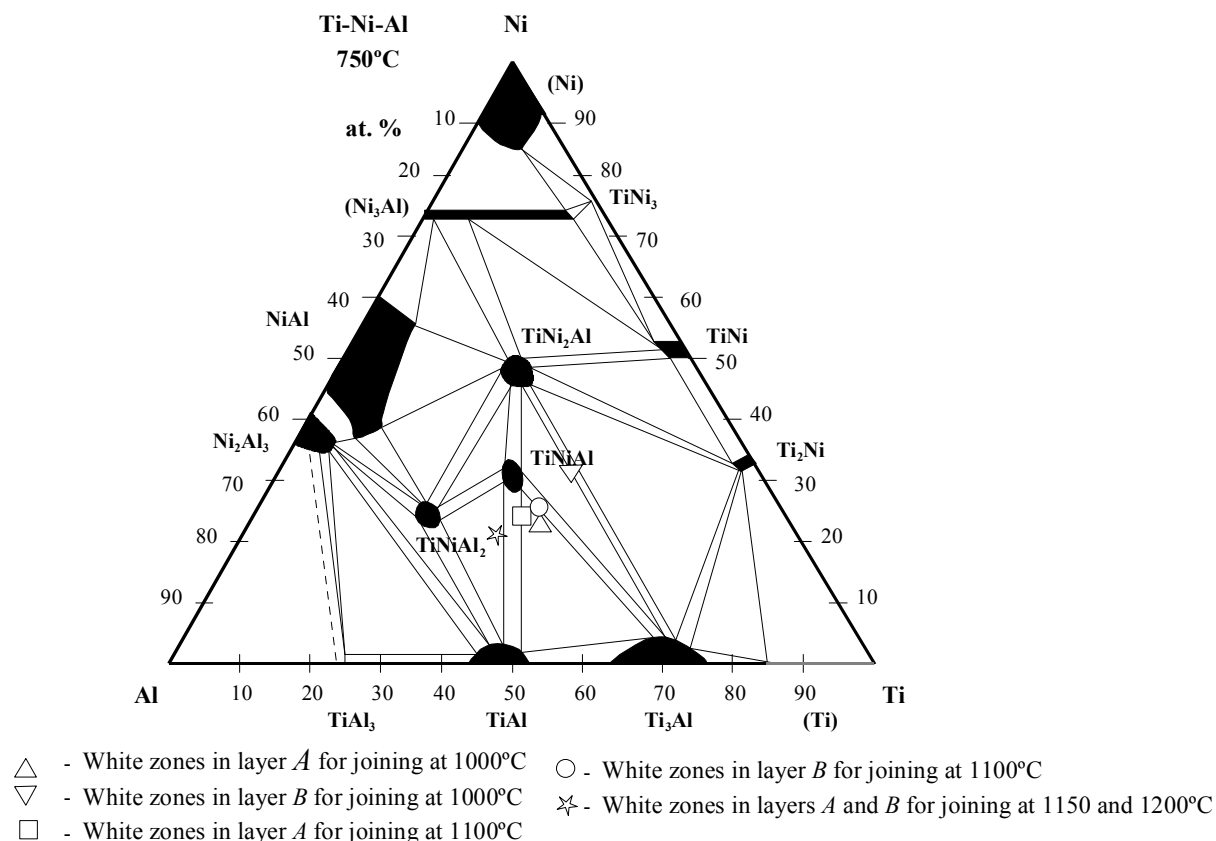


Fig. 3 – Isothermal section at 750°C of the ternary Ti-Ni-Al phase diagram [10], where the composition of some of the reaction products detected at the interfaces are marked.

Based on the results obtained in this investigation, we think that the reaction mechanism underlying the formation of the interface is as follows: In the course of the heating to the brazing temperature, a liquid phase is formed in the Tini67 foil, due to the interdiffusion of Ti and Ni atoms within the braze alloy. When the liquid reaches the γ -TiAl alloy surface it begins to dissolve it, incorporating Ti, Al, Nb and Cr. The absence of Ni in the region of the base alloy close to the interface and the presence of Al in high contents throughout all the joint allow us to conclude that dissolution of the base intermetallic by the braze alloy plays, in this case, a major role than does diffusion of the braze alloy elements into the base intermetallic.

When the solubility limits into the melt are exceeded, several intermetallic compounds begin to precipitate. Ti and Al atoms will combine preferentially, originating the formation of α_2 -Ti₃Al (D₁ and D₂ dark zones). Since the α_2 phase dissolves practically no Ni (see Fig. 3), most of the Ni will be segregated and will combine with the remaining Ti and Al atoms to form TiNiAl (white zones). Near the γ -TiAl surface, a reaction layer (layer A) composed of an intricate mixture of α_2 -Ti₃Al and TiNiAl is formed. A thin and quasi-continuous border, consisting essentially in α_2 -Ti₃Al delimits this layer from the central zones of the interface (layer B) that consists in a matrix composed of TiNiAl, in which coarse α_2 -Ti₃Al particles are dispersed.

Attending to the elemental distribution profiles and the evolution of Ti:Al atomic ratios across the interface, when the Ti contents reaches the sharp maximum one verifies that: (1) the Al contents reaches the minimum value; (2) the Ti:Al ratio attain the maximum, which is close to 3.0; (3) the corresponding zone of the interface is the quasi-continuous thin border, consisting of α_2 -Ti₃Al, that delimits layer A from layer B. Therefore, we suppose that when the value of the Ti:Al atomic ratio is close to 3.0, the driving force for the formation of α_2 -Ti₃Al is more intense, and this compound

precipitates preferentially to any other at the locus where this relation is verified; this is what should happen along the zone corresponding to thin border that delimits layers *A* and *B*.

As the brazing temperature is raised the solubility limits of the different elements into the melt will be higher, the dissolution of the γ -TiAl alloy will be more intense, the resulting interfaces will be richer in Ti, Al, Cr and Nb, and wider too. Further, the slope of distribution profiles of Ti and Al will be smoother and the locus along the interface where the value of Ti:Al ratio reaches 3.0 will be located farther away from the γ -TiAl surface; consequently the thickness of layer *A* will increase.

The α_2 -Ti₃Al zones will grow significantly, namely those located in layer *A*, and the segregation of Ni becomes evident when joints are processed at 1150°C: although, in layer *A*, TiNiAl zones are still detected at some α_2 -Ti₃Al grain boundaries, most of them are agglomerated near the border delimiting this layer from layer *B*. As a result of the increasing dissolution of the γ -TiAl alloy, when brazing is performed at 1150 and 1200°C, the concentration of Al will be high enough to allow, at some zones of layer *A*, the formation of γ -TiAl (*D*₃ zones).

Conclusions

The joining of Ti-47Al-2Cr-2Nb in the temperature range of 1000 to 1200°C using Tini67 as braze alloy, produces interfaces consisting in two distinct layers (*A* and *B*) that should be essentially composed of α_2 -Ti₃Al and TiNiAl intermetallics. TiNiAl is mainly found at the central zones (layer *B*) of the interfaces and near the thin border that delimits these zones from the reaction layer (layer *A*) formed close to the γ -TiAl alloy. For joining at 1150 and 1200°C, a phase that should consist of γ -TiAl is also detected in layer *A*. The increase of the extension of the interface when the brazing temperature rises is mainly due to the growth of layer *A*. The dissolution of the γ -TiAl by the braze and the diffusion of the γ -TiAl elements into joint area appear to play the leading roles on the formation of the interfaces.

References

- [1] D.M. Dimiduk: Materials Science and Engineering A263 (1999), p.281.
- [2] N.S. Stoloff, C.T. Liu and S.C. Deevi: Intermetallics 8 (2000), p.1313.
- [3] E.A. Loria: Intermetallics 8 (2000), p.1339.
- [4] E.A. Loria: Intermetallics 9 (2001), p.997.
- [5] Q. Xu, M.C. Chaturvedi, N. L. Richards and N. Goel: Structural Intermetallics 1997, The Minerals, Metals & Materials Society (1997), p. 323.
- [6] G. Çam and M. Koçak: J. Mater. Science 34 (1999), p.3345.
- [7] H. Holmquist, V. Recina and B. Pettersson: Acta Mater. 47 (1999), p.1791.
- [8] W. Glatz and H. Clemens: Intermetallics 5 (1997), p.423.
- [9] Binary Alloy Phase Diagrams CD-ROM, 2nd Edition plus updates (ASM International, 1996).
- [10] Handbook of Ternary Alloy Phase Diagrams (ASM International, 1994).