Assessing the Influence of Heat Treatments on γ-TiAl Joints

A. Guedes(1), a, A. M. P. Pinto(1), b, M. F. Vieira(2), c, F. Viana(2), d

(1)Universidade do Minho, Departamento de Engenharia Mecânica, CIICS, Portugal

(2)Faculdade de Engenharia da Universidade do Porto, Departamento de Engenharia
Metalúrgica e Materiais, GMM/IMAT, Portugal

aaguedes@dem.uminho.pt, banapinto@dem.uminho.pt, cmvieira@fe.up.pt,
dfviana@fe.up.pt

Abstract. The heat treatment of γ-TiAl alloy (Ti-47Al-2Cr-2Nb (at.%)) diffusion brazed joints was investigated. Joining was performed using a Ti/Ni/Ti clad-laminated braze alloy foil at 1050 and 1150ºC with a dwell time of 10 minutes. The joints were subsequently heat treated at 1250 and 1350ºC for 240 and 30 minutes, respectively. The microstructure and the chemical composition of the interfaces were analysed by scanning electron microscopy (SEM) and by energy dispersive X-ray spectroscopy (EDS), respectively. Microhardness tests performed across the interface were used to roughly predict the mechanical behaviour of the as-diffusion brazed and of the heat treated joints. Diffusion brazing produced interfaces with two distinct layers essentially composed of α_2-Ti$_3$Al and of TiNiAl; γ-TiAl was also detected for joining at 1150ºC. After heat treating, the as-diffusion brazed microstructure of the interface was completely replaced by a mixture essentially composed of γ-TiAl and α_2–Ti$_3$Al single phase grains and of (α_2 + γ) lamellar grains. Microhardness tests showed that the hardness of the as-diffusion brazed interfaces, which ranges from 567 to 844 HV (15 gf), is significantly higher than that of the titanium aluminide alloy (272 HV). All post-joining heat treatments lowered substantially the hardness of the interface, as the hardness of the main phases detected at the interfacial zone after heat treating the joints is comprised between 296 and 414 HV.

Introduction

Titanium aluminide alloys based on the intermetallic compound γ-TiAl are promising candidates for high performance applications on aircraft engine components, aerospace vehicles and automotive engines, owing to their low density ($3,7 - 4,0$ g.cm^{-3}) high specific stiffness, excellent strength retention at high temperature and good creep and oxidation resistance [1,2,3]. More recently the potential possibility of using γ-TiAl alloys as biomaterials for endoprothesic applications has been investigated. In vitro experiments showed that pre-oxidised Ti-45Al-2W-0.6Si-0.7B is better than the Ti6Al4V alloy currently used as biomaterial [4].

Joining is a key technology for the practical application of γ-TiAl alloys. Brazing and diffusion brazing are straightforward joining routes to produce γ-TiAl joints [5-9]. In our previous investigations Tini 67 [10,11] and Ticuni [11-13] were found to be adequate fillers to produce γ-TiAl diffusion brazed joints in the temperature range of 1000 to 1200ºC. In comparison to fusion welding processes this technique does not induce high residual thermal stresses at the interface, which in conjunction with the formation of brittle phases at the interface are the main causes of solid state cracking of joints. Additionally, diffusion brazing does not require the application of elevated bonding pressures, as does solid state diffusion bonding, and contrarily to the latter it can be successfully applied to nearly all joint geometry. Avoiding or decreasing the formation of brittle phases at interface is actually one of the main concerns of the investigation on the joining by diffusion brazing of γ-TiAl alloys. This could be accomplished either by (1) selecting adequately the chemistry of the filler, but it often results on the formation of soft reaction products that are not

able to withstand the temperatures at which γ-TiAl alloys are projected to operate as structural materials for engine components, or by (2) performing a post-joining heat treatment in order to dissolve total or partially the undesirable brittle compounds formed at the interface. Unfortunately heat treating the joints is not always an industrially viable procedure as it may require extremely long dwell times (several days) and elevated temperatures [5], which raise dramatically the final cost of the joining process.

The present investigation is focused on the effect of post-joining heat treatments on the microstructure and hardness of γ-TiAl alloy diffusion brazed joints, using a Ti-Ni-Ti clad-laminated alloy foil (Tini 67) as filler. The dissolution of the majority of the "undesirable" brittle phases formed at the as-diffusion brazed interface was obtained after heat treating using substantially shorter dwell times than those reported in the literature.

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, which consists of a mixture of γ-TiAl grains and lamellar (α_2-Ti$_3$Al + γ-TiAl) grains. A Tini 67 foil (71.3Ti-28.7Ni, at.%) with a thickness of 50 μm was selected as filler alloy. Tini 67 is a clad-laminated alloy produced by Wesgo Metals, whose solidus and liquidus temperatures are 942 and 980ºC, respectively. This braze alloy consists of a 12 μm Ni foil sandwiched between two 19 μm Ti foils.

Diffusion brazing was performed at 1050 and 1150ºC with a dwell time of 10 minutes at the joining temperature. The details of the joining procedure are described elsewhere [10,11]. The joints brazed at 1050 were heat treated at 1350ºC for 30 minutes; those obtained for joining at 1150ºC were heat treated at 1250 and 1350ºC for 240 and 30 minutes, respectively. Heat treatments were performed in vacuum in an electrical furnace. The furnace chamber was evacuated by a combination of rotary and turbomolecular pumps to a vacuum level that remained better than 8 x 10^{-5} mbar during the entire thermal cycle. The heating and cooling rates were fixed at 3ºC.min^{-1} by a temperature controller.

Samples for microstructural and chemical characterisation and for microhardness tests were cut perpendicularly to the interface; cold mounted in epoxy resin and prepared using standard metallographic techniques. The interfaces were examined by SEM and chemically analysed by EDS at an accelerating voltage of 15 keV. Vickers microhardness tests were performed throughout the interface. The load (15 gf) was applied in 15 s with a dwell time of 15 s.

Results and discussion

The reaction between the Tini 67 foil and the γ-TiAl alloy induced the formation of interfaces, apparently free of pores and cracks, essentially composed of Ti, Al and Ni that could be divided into two distinct reaction layers (labelled as A and B) as shown in figure 1. No perceptible alterations were detected either on the chemical composition or on the microstructure of the γ-TiAl alloy after joining. Attending to the results of the EDS analysis performed across the interface [10,11], in conjunction with the information supplied by the Ti-Al [14] and the Al-Ni-Ti [15] phase diagrams, the main phases formed at the interfaces are the intermetallic compounds α_2-Ti$_3$Al and TiNiAl; a small amount of γ-TiAl was also detected for joining at 1150ºC (see figure 1). A detailed discussion of the nature of the phases formed at the interface is done elsewhere [10,11]. It should be noted that all of these phases have been identified by X-ray diffraction analysis performed on the fracture surfaces of shear tested samples [11].

The microstructure of the interface after heat treating as well as the results of the EDS analysis performed across the interface of the heat treated joints are presented in figure 2 and in table 1, respectively. Additionally, the compositional plots of the EDS analysis results after heat treating are marked on the Al-Ni-Ti isothermal section at 750ºC shown in figure 3. First it should be noted that the resulting microstructure of the interface is mainly determined by the heat treatment conditions, i.e., if joints are processed at different temperatures and subsequently subjected to the same heat

treatment, then the resulting interfacial microstructure is the same. Secondly, none of the heat treatments produced noticeable changes either on the microstructure or on the chemical composition of the γ-TiAl alloy. The main effects of both heat treatments on the features of the interface could be summarized as follows:

(1) The extension of the interface increases significantly, from about 110/120 μm in the as-diffusion brazed condition, to approximately 545/610 μm after heat treating;

(2) The as-diffusion brazed microstructure of the interface is completely replaced by a mixture mainly composed of α_2–Ti_3Al (zones 3 and 4 in Fig. 2(b) and Table 1) and γ-TiAl (zone 2) single phase grains and of ($\alpha_2 + \gamma$) lamellar grains (zone 1); most of the TiNiAl phase was dissolved (small particles, as the ones identified as zone 5, are still present);

(3) The chemical composition of the phases detected at the interface after heat treating is independent of the heat treating temperature;

(4) An extensive formation of γ-TiAl (only detected for joining at 1150ºC) as well as of a ($\alpha_2 + \gamma$) lamellar constituent (not detected after joining) is observed throughout the interface.
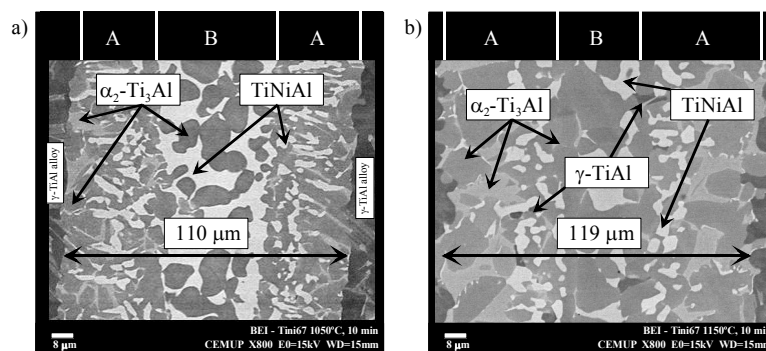

Fig. 1 – Backscattered electron images (BEI) of the interface, obtained after joining at:
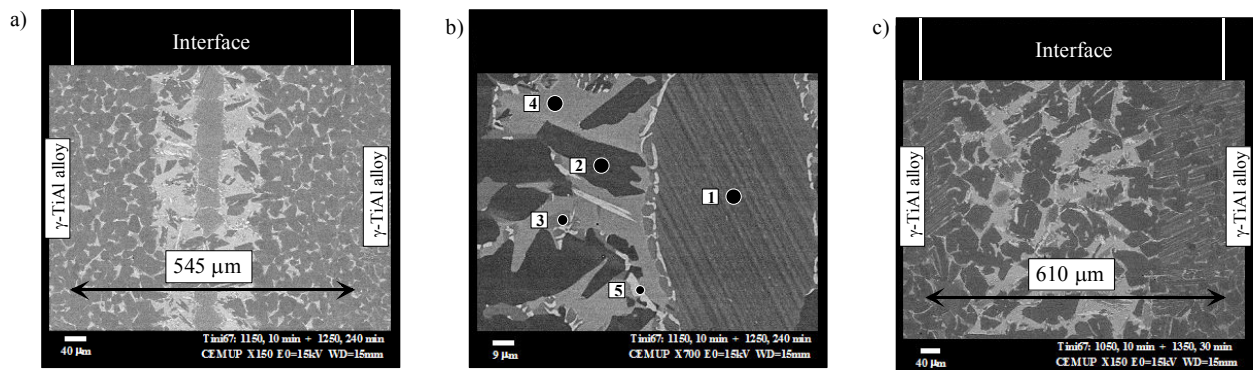a) 1050ºC; b) 1150ºC.

Fig. 2 – Backscattered electron images (BEI) of the interface, obtained after heat treating at:
a) and b) 1250ºC for 240min.; c) 1350ºC for 30min.. b) is a detail of the central region of the interface.

Table 1 – EDS chemical analysis results of the specimens after heat treating.

Temperature, dwell time (ºC, min)	Zone	Ti	Al	Ni	Cr	Nb	Ti/Al	Possible phase(s)
				(at.%)				
1250, 240 1350, 30	1	54.5	41.2	1.0	1.5	1.8		α_2-Ti_3Al + γ-TiAl
	2	52.4	44.2	0.9	0.8	1.7	1.2	γ-TiAl
	3	62.2	34.5	0.3	1.5	1.5	1.8	α_2-Ti_3Al
	4	60.0	31.3	2.0	4.7	2.0	1.9	α_2-Ti_3Al
	5	36.1	39.9	18.6	3.8	1.6		TiNiAl

Microhardness measurements showed that the interface is significantly harder than the γ-TiAl alloy (272 ± 29 HV), since the hardness of the main phases detected at the interfaces is comprised, between 567 ± 34 (α₂-Ti₃Al, for joining at 1050ºC) and 844 ± 20 HV (TiNiAl, for joining at 1150ºC). Both heat treatments lowered substantially the hardness of the interface. The hardness of the main phases detected at the interface after heat treating the joints is comprised between 296 ± 35 (γ-TiAl) and 414 ± 32 HV (α₂–Ti₃Al).A comparison between the hardness of the main phases that constitute the as-brazed interfaces and those detected after heat treating the joints is presented in table 2.
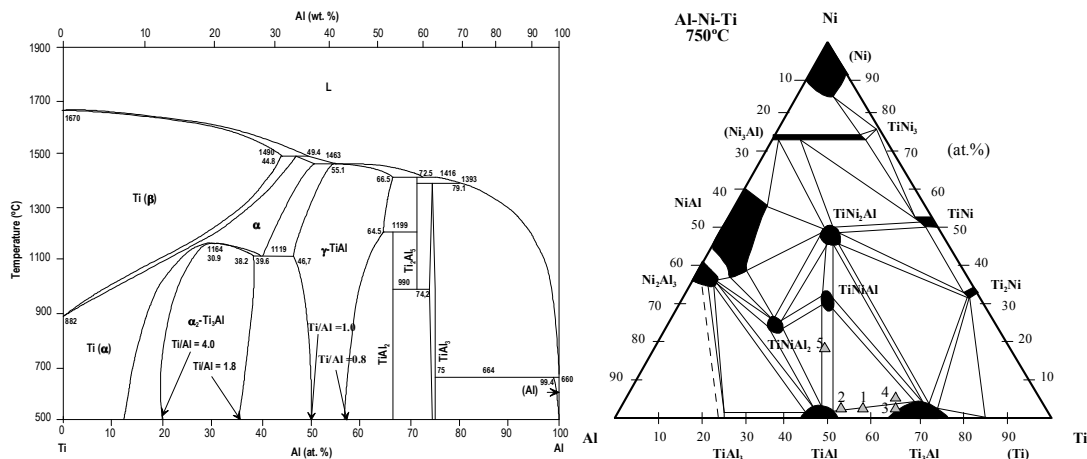

Fig. 3 – Ti-Al phase diagram [14] and Al-Ni-Ti isothermal section at 750ºC [15] where are marked the compositional plots of the EDS analysis results after heat treating.

Table 2 - Hardness values (HV 0.015) of the γ-TiAl alloy and of the main phases/constituents detected at the interface in the as-diffusion brazed condition and after heat treating the joints.

Possible phases/constituents	Diffusion brazing temperature (ºC)		Post-joining heat treatment (1250 or 1350ºC)
	1050	1150	
α₂–Ti₃Al	567 *(34)*	570 *(38)*	414 *(32)*
γ-TiAl	-	-	296 *(35)*
TiNiAl	594 *(39)*	844 *(20)*	-
Lamellar constituent (α₂ + γ)	-	-	385 *(32)*
γ-TiAl alloy	272 *(29)*		

Standard deviation in parenthesis

The results obtained in this investigation, in conjunction with the information supplied by the Ti-Al and the Al-Ni-Ti equilibrium phase diagrams; allow inferring some of the transformations induced by the heat treatments. The Ti-Al phase diagram (Fig. 3) shows that the α₂–Ti₃Al phase is not stable at 1250ºC neither at 1350ºC. On heating to the heat treating temperature, this phase undergoes one of the following sequences of transformations: α₂-Ti₃Al → Ti (α) (when heated to 1250ºC) and α₂-Ti₃Al → Ti (α) → Ti (β) (when heated to 1350ºC); the inverse sequences occur when cooling to room temperature. Taking into account the Al-Ni-Ti ternary phase diagram (see Fig. 4), TiNiAl melts bellow 1250ºC and the resulting liquid coexists in equilibrium with γ-TiAl and a Ti solid solution. We presume that the liquid ensuing from the melting of TiNiAl will dissolve part of the Ti (α) (for heat treating at 1250ºC) or part of the Ti (β) (for heat treating at 1350ºC) until the solubility limits are reached. Then, γ-TiAl and Ti (α) or Ti (β) begin to precipitate from the liquid. The Ti solid solution will transform into α₂-Ti₃Al upon cooling to room temperature while the remaining liquid originates TiNiAl. Finally, the α₂ + γ lamellar constituent should correspond to the loci of the interface where the undissolved Ti solid solution has a

composition close to that of the Ti-Al eutectoid at 1119ºC; upon cooling Ti (α) will transform into the $\alpha_2 + \gamma$ lamellar constituent as a result of the eutectoid reaction.
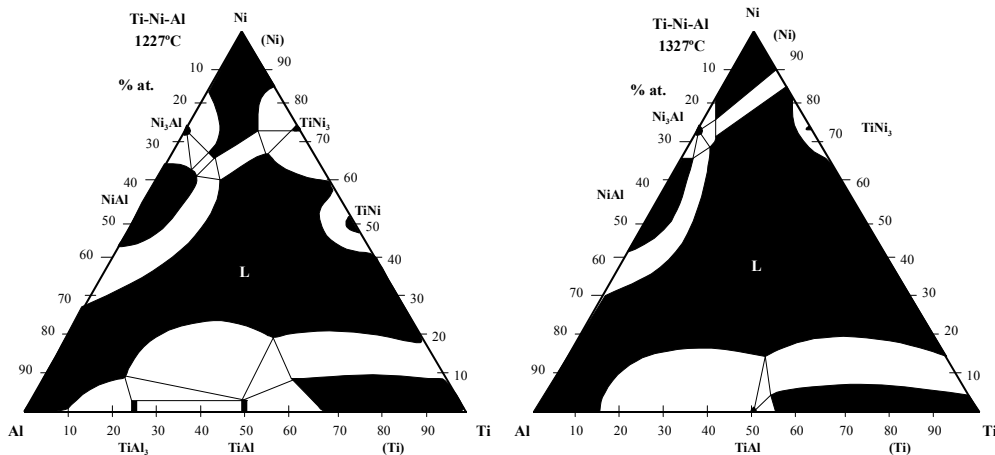

Fig. 4 – Isothermal sections through the Al-Ni-Ti phase diagram at 1227 and 1327ºC [15].

Conclusions

Post-joining heat treatments performed at 1250 and 1350ºC with dwell times of 240 and 30 minutes, respectively, modify radically the microstructure of joints: the main constituents detected at the interface evolve from α_2-Ti$_3$Al and TiNiAl (in the as-brazed condition) to γ-TiAl and α_2-Ti$_3$Al after heat treatment. Most of the TiNiAl compound is dissolved; only very small particles are still present.

The hardness of the interface decreases with the post-joining heat treatments, from values comprised between 567 and 844 HV, to values comprised between 296 and 414 HV. This is essentially the result of the combined effect of both the dissolution of TiNiAl (hardest constituent) and the extensive formation of γ-TiAl (softer constituent). However, the hardness of interface still remains higher than that of the intermetallic alloy (272 HV).

References

[1] E.A. Loria: Intermetallics 9 (2001), p. 997.

[2] S. Djanarthany J, J-C. Viala, J. Bouix: Mater. Chem. Phys. 72 (2001), p. 301.

[3] N.S. Stoloff, C.T. Liu, S.C. Deevi: Intermetallics 8 (2000), p. 1313.

[4] M.L. Escudero, M.A. Munoz-Morris, M.C García-Alonso, E. Fernández-Escalante: Intermetallics 12 (2004), p. 253.

[5] Q. Xu, M. C. Chaturvedi, N. L. Richards, N. Goel: Structural Intermetallics 1997, The Minerals, Metals & Materials Society (1997), p. 323.

[6] R.K. Shiue, S.K. Wu, S.Y. Chen: Acta mater. 51 (2003), p. 1991.

[7] S.J. Lee, S.K. Wu: Intermetallics 7 (1999), p. 11.

[8] R.K. Shiue, S.K. Wu, S.Y. Chen: Intermetallics 12 (2004), p. 929.

[9] I.C. Wallis, H.S. Ubhi, M.-P. Bacos, P. Josso, J. Lindqvist, D. Lundstrom, A. Wisbey: Intermetallics 12 (2004), p. 303.

[10] A. Guedes, A.M.P. Pinto, M. Vieira, F. Viana: Mater. Sci. Forum, Vol. 455-456 (2004), p. 880.

[11] A. Guedes: PhD thesis, Universidade do Minho (2004).

[12] A. Guedes, A.M.P. Pinto, M. Vieira, F. Viana: Mater. Sci. Forum, Vol. 426-432 (2003), p. 4159.

[13] A. Guedes, A.M.P. Pinto, M. Vieira, F. Viana: J. Mater. Sci. 38 (2003), p. 2409.

[14] Binary Alloy Phase Diagrams CD-ROM, 2nd Edition plus updates, ASM International, 1996.

[15] P. Villars, Handbook of Ternary Alloy Phase Diagrams, ASM International, 1994.