World Applied Sciences Journal 15 (11): 1507-1511, 2011
ISSN 1818-4952

TLP Bonding of a Gamma Prime Strengthened Superalloy Using Ni-Si-B Interlayer at 1150°C-Part II: Mechanical Properties

M. Pouranvari

Young Researchers Club, Dezful Branch, Islamic Azad University, Dezful, Iran

Abstract: In this paper the effect of transient liquid phase (TLP) bonding condition on the mechanical properties of GTD-111 nickel base superalloy investigated. Shear strength and hardness profile of the joints were discussed with respect to the bond microstructure. In the bonding condition, in which isothermal solidification has not been completely accomplished, eutectic constituent which has the highest hardness in the bond region is the preferential failure source. At the bonding time of 45 min at 1150°C, when the eutectic products are completely removed, bonds with shear strength of about 70% of that of the base metal are achieved. After post bond heat treatment at 1150°C for 240 min, significant γ′ phase, formed within the bond region, increased the bond shear strength. The shear strength of the homogenized bond was about 90% of that of the base metal.

Key words: TLP bonding · Nickel based superalloy · Isothermal solidification · Mechanical properties

INTRODUCTION

Two important key microstructural requirements of γ′ strengthened nickel base superalloys joints are as follows:

- The avoidance of undesired intermetallic phase in the middle of the joint.
- Development of a desired γ/γ′ microstructure in the joint region.

Production of joints which fulfill the microstructural requirements for high stresses and temperatures can be achieved *via* transient liquid phase (TLP) bonding or so called diffusion brazing process [1-7]. In general, it is considered that there are three distinct stages during diffusion brazing, namely: base metal dissolution, isothermal solidification and solid-state homogenization. Combining isothermal solidification with a subsequent solid state homogenization treatment, offers the possibility of producing ideal joints [8].

A typical microstructure of TLP bonded joint of a nickel based precipitation hardened superalloy such as GTD-111 using a boron containing interlayer, consists of three distinct microstructural zones, before completion of isothermal solidification [2]:

- Athermally Solidified Zone (ASZ) which usually consists of eutectic microconstituents. This zone is formed due to insufficient time for isothermal solidification completion. Cooling is the main driving force for athermal solidification (i.e. non-isothermal solidification).
- Isothermally Solidified Zone (ISZ) which usually consists of a solid solution phase. Compositional change induced by interdiffusion between substrate and interlayer during holding at a constant bonding temperature is the driving force for isothermal solidification. As a result of the absence of solute rejection at the solid/liquid interface during isothermal solidification under equilibrium, formation of second phase is basically prevented [3].
- Diffusion Affected Zone (DAZ) which consists of boride precipitates due to B diffusion into the base metal (BM) during TLP bonding.

The presence of different zones of TLP bonded joint is schematically shown in Fig. 1.

The aim of this research was to investigate the effect of isothermal solidification and post bond heat treatment on the mechanical properties of the TLP bonded GTD-111 nickel based superalloy.

Corresponding Author: M. Pouranvari, Young Researchers Club, Dezful Branch, Islamic Azad University, Dezful, Iran.
Tel: +989124075960, Fax: +98 21 88522421.

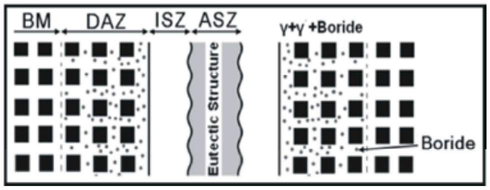


Fig. 1: Schematic representation of various microstructural zones in a TLP bonded γ′ strengthened nickel based superalloys before isothermal solidification completion

EXPERIMENTAL PROCEDURE

GTD-111 superalloy was used in the standard heat treatment condition as the base metal in this investigation. Also, a commercial Ni-Si-B alloy (MBF30), in the form of an amorphous foil with 25.4 µm thickness was used as the interlayer. The detailed experimental set up is given in the Part I [9]. TLP bonding operation was carried out in a vacuum furnace under a vacuum of approximately 10^{-4} Torr. Bonding temperature of 1150°C was chosen and bonding time varied from 30 to 45 min. Complete isothermally solidified bonds were homogenized at 1150°C for 240 min in an argon gas atmosphere (%99.999 Ar) using a tunnel furnace.

For metallurgical investigation, the bonded specimens were sectioned perpendicularly to the bond. Microhardness test, a technique that has proven to be useful in quantifying microstructure-mechanical property relationships, was used to determine the joint region hardness profile. The test was conducted on sample cross section using a 25 g load on a Buehler microhardness tester.

For mechanical testing, the shear test was chosen. Unlike the tensile testing, during the shear test the boding region is essentially stressed. Therefore, the shear testing is a more appropriate test to evaluate the mechanical properties of the TLP bonds compared to the tensile testing. Room temperature shearing test was conducted according to ASTM D1002-05 standard using an Instron tensile machine with a cross-head speed of 2 mm/min. Before shearing test, the edge effects were removed by machining. The shear fixture used for testing is shown schematically in Figure 2. This fixture subjects the sample to a pure shear stress at the bond line. A metal sleeve was placed over the test jig in order to prevent movement of specimens. For comparison, shear tests was also conducted on parent alloy.

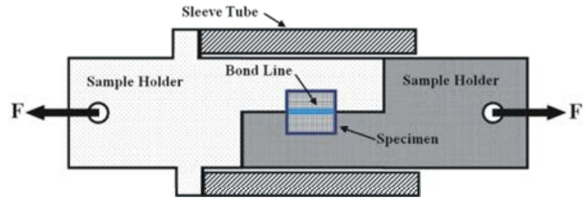


Fig. 2: Schematic of Shear test fixture

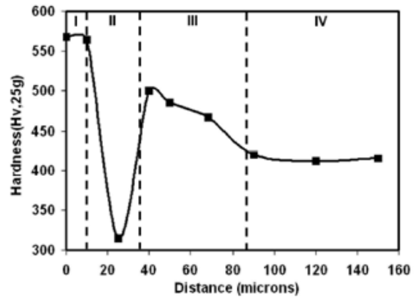
RESULTS AND DISCUSSION

Hardness Characteristics of the TLP Bonds: Hardness profile across the joint region is a quantitative mechanical properties measurement of different zones in the joint region. Hardness profile is a good indicator of bond microstructure and can be used to assess the effect of secondary phase precipitates on mechanical properties. It can also used for assessing degree of homogenization.

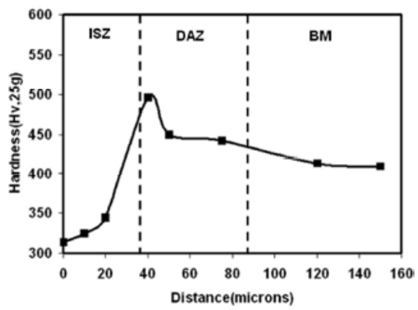
Figure 3a shows hardness profile of bonds made at 1150°C for 30 min indicating four distinct zones:

- Region I corresponds to the ASZ. According to the microstructure of ASZ, the peak hardness in this zone is due to the eutectic type structure which contains hard brittle nickel boride. This region provides an easy crack propagation path. Hence it is necessary to eliminate this eutectic structure in order to improve the strength of TLP joints.
- Region II corresponds to ISZ which has lower hardness relative to the base metal. Extent of interdiffusion between interlayer and substrate determine the hardness of ISZ. Low hardness of ISZ can be related to insufficient diffusion of alloying elements such as solid solution strengthening elements (eg, Co) and γ′ forming elements, Al and Ti.
- Region III corresponds to DAZ. Hardness peak which observed in this region can be related to carbo-boride precipitates.
- Region IV corresponds to base metal.

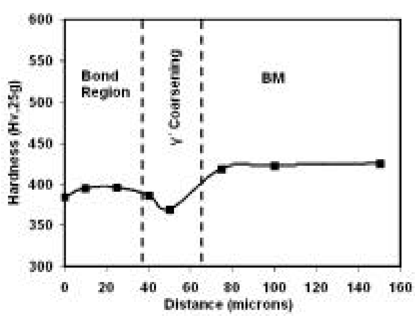
Hardness profile across the joint region of bonds made at 1150°C for 45 min is shown in Figure 3b. At this bonding condition complete isothermal solidification has been. As can be seen from Figure 3b, the ASZ hardness decreases to the values of ISZ, but the peak in the hardness profile exist for region III (DAZ) due to presence of boride precipitates. This is due to the fact that these precipitates are not formed due to non-isothermal solidification. Indeed, these borides precipitates are



(a)



(b)



(C)

Fig. 3: Hardness profile across the a) bond made at 1150°C for 30 min b) bond made at 1150°C for 45 min and c) bond made at 1150°C and then homogenized

formed due to solid-state diffusion of B into the BM during isothermal holding of the samples in the bonding temperature. Therefore, the completion of isothermal solidification has not influence on the boride precipitates in diffusion affected zone. Results of hardness testing indicated low hardness of the joint's centerline, which can be due to the formation of insufficient gamma-prime within the joint, leaving a relatively soft bond region. Indeed, Extent of interdiffusion between interlayer and substrate determine hardness of ISZ. As can be seen from Figure 3b hardness continuously increases from the bond centerline toward bond line due to the diffusion of alloying element from base metal to this region.

Hardness profile of bonds made at 1150°C for 45 min and then homogenized at 1150°C for 240 min is shown in Figure 3c. As can be seen, hardness profile across the bond region was affected significantly by PBHT. Hardness of joint region depends on size and volume fraction of γ′ precipitates and concentration of solid solution elements. This figure shows that bond region hardness increases in homogenized bonds. This can be related to the formation of significant γ′ at the bond region due to the diffusion of Al and Ti from base material into the bond region during PBHT. However, hardness values are not constant within the joint region because the γ′ volume fraction across the joint is not constant and shows slight variations towards the middle of the joint. As seen in Part I [9] post bond heat treatment removes the carbo-boride precipitates in DAZ causing the significant decrease in the hardness of this region. However a narrow soft zone exists in the base metal adjacent to the joint interface. As was documented in Part I [9] microstructure of transition zone between the bond region and base metal exhibit coarsened gamma prime precipitates. Lower hardness of this zone can be attributed to γ′ coarsening. Gamma prime coarsening adjacent to the bond region is also has been reported by Schnell [10] in TLP bonding of single crystal nickel base superalloy CMSX-4 using a Ni-Cr-Co-Al-Ta-B filler alloy. The γ′ coarsening can be related partly to the formation of carbo-borides in DAZ and partly to the experienced thermal cycle of PBHT [11]. It is reported that that boron diffusion induced precipitates in DAZ can significantly change the morphology and size of γ′ [11, 12].

Shear Strength of TLP Bonds: Figure 4 shows the results of shear tests on athermally solidified bond, isothermally solidified bond and post bond heat treated. For comparison, shear strength of parent metal were also plotted. As can be seen, shear strength of the bonds made at 1150°C for 30 min is the lowest. In this bonding condition, the joint's centerline consists of athermal solidification product. High hardness of eutectic structure (Figure 3a) coupled with the fact that nickel boride phase form interlinked network provide a metallurgical notch which decreases significantly load carrying capacity of the joint. This has been verified by fractography of fracture surfaces. SEM micrographs of fracture surface of bond made at 1150°C for 30 min are shown in Figure 5. Semi-cleavage/ intergranular morphology of fracture surface of this joint which are low energy fracture modes [13], confirms low shear strength of this bond.

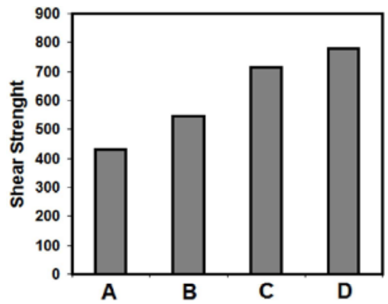


Fig. 4: Comparison of shear strength of bonds and base metal: (A) Athermally solidified bond, (B) isothermally solidified bond, (C) Isothermally solidified bond and then homogenized (D) Base metal

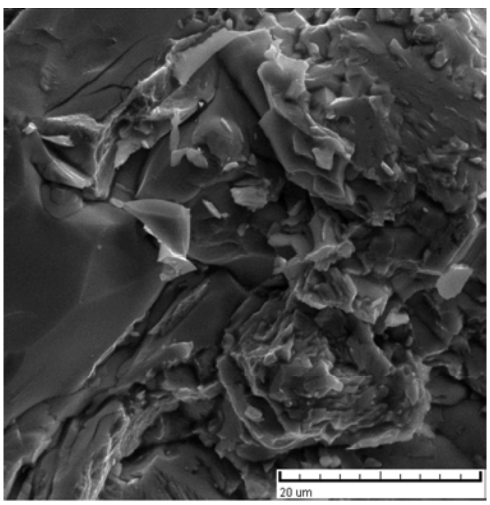


Fig. 5: SEM micrograph showing fracture surface of bonds made at 1150°C for 30 min

As can be seen in Figure 4, At 45 min holding time at 1150°C, in which eutectic structure was completely removed, bonds with shear strength of about 70% that of the base metal were achieved. It can be deduced that in bonding condition with the complete isothermal solidification, the amount of γ′ precipitates in the bond region is the dominant factor for the shear strength.

The results show that homogenized bonds have the highest shear strength, which is about 90% that of the parent alloy shear strength. The increase in shear strength of homogenized bonds can be attributed to the following reason:

- Develop more homogenize microstructure across the bond region, as confirmed by hardness profile.
- The increase of γ′ volume fraction at the bond region after PBHT due to more diffusion time for Ti and Al.

CONCLUSIONS

From this research the following conclusions can be drawn:

- In bonding condition in which isothermal solidification is not completely accomplished, eutectic type structure of ASZ which has the highest hardness in the bond region is the preferential failure source.
- At the bonding time of 45 min at 1150°C, when the eutectic products are completely removed, bonds with shear strength of about 68% of that of the base metal are achieved.
- After PBHT, significant γ′ phase, formed within the bond region, increased the bond shear strength. The shear strength of the homogenized bond was about 90% of that of the base metal.
- The hardness profile of TLP joint became more uniform after post bond heat treatment, due to the formation of significant amount of γ′ precipitates in the bond region and the removal of carboboride precipitates from DAZ. A soft region was observed adjacent to the joint interface, due to γ′ coarsening in this region, which may contribute to the reduction of joint shear strength. Therefore, considering the softening effect of γ′ coarsening zone, there is a need to design a proper solution/aging treatment for the homogenization of the microstructure across the joint.

REFERENCES

1. Duvall, D.S., W.A. Owczarski and D.F. Paulonis, 1974. Weld. J., 53: 203-214.
2. Pouranvari, M., 2010. J. Metallurgy-MJoM, 16: 241-247.
3. Pouranvari, M., 2010. J. Metallurgy-MJoM, 16: 249-254.
4. Pouranvari, M., 2010. J. Metallurgy-MJoM, 17: 87-94.
5. Pouranvari, M., A. Ekrami and A.H. Kokabi, 2008. J. Alloys Compd, 461: 641-647.
6. Ojo, O.A., N.L. Richards and M.C. Charturvedi, 2004. Sci. Technol.Weld. Joining, 9: 532-540.
7. Liu, J.D., T. Jin, N.R. Zhao, Z.H. Wang, X.F. Sun, H.R. Guan and Z.Q. Hu, 2007. Sci. Technol. Weld. Joining, 12: 381-385.
8. Gale, W.F. and D.A. Butts, 2004. Sci. Technol. Weld. Joining, 9: 283-300.

9. Pouranvari, M., 0000. Middle East J. Scientific Research, Submitted, Part I.
10. Schnell, A., 2004. A study of the diffusion brazing process applied to the single crystal superalloy CMSX-4, PhD Thesis, Institut Des Matériaux.
11. Pouranvari, M., A. Ekrami and A.H. Kokabi, 2008. Mater. Sci. Eng. A., 490: 229-234.
12. Idowu, O.A., O.A. Ojo, M.C. Chaturvedi, 2006. Metall. Matter. Trans A., 37: 2787-96.
13. Hertzberg, W., 1996. Deformation and fracture mechanics of engineering materials, New York, John Wiley and Sons.