



MICROSTRUCTURAL DEVELOPMENT AND MECHANICAL PROPERTY RELATIONSHIPS IN WIDE-GAP TRANSIENT LIQUID PHASE BONDING OF SINGLE CRYSTAL TO POLYCRYSTALLINE NICKEL-BASE SUPERALLOYS

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ABSTRACT

This paper investigates the microstructure-mechanical property relationships in wide gap transient liquid phase (TLP) bonds, between the single crystal nickel-base superalloy CMSX-4 and two polycrystalline superalloys, IN 738 and IN 939, using wide-gap style composite interlayers of Niflex-110 and Niflex-115, in comparison with conventional BNi-3 foils. Composite interlayers resulted in the suppression of bondline boride formation. Results from shear testing and fractography indicated that ductile shear failure occurred along the bondline and that the extent of formation of γ' on the bondline was the dominant factor in the determination of room-temperature shear strength of the bonds.

1. INTRODUCTION

Bonds between single crystal and polycrystalline nickel-base superalloys are of interest for gas turbine applications, in both aerospace and especially land-based power generation. Single crystal nickel-base superalloy turbine blades are well known for their superior high temperature creep properties. The interest in joining dissimilar (single crystal to polycrystal) nickel-base superalloys arises primarily from the difficulty in producing large single crystal blades for power generation applications.

Fusion welding, diffusion bonding and brazing are the three main joining techniques used in the industry for joining high temperature materials [1]. Conventional fusion welding, such as GTA suffers from hot and post-weld heat treatment cracking and sometimes formation of brittle phases in the heat affected zone (HAZ), when welding nickel-base superalloys [2]. Diffusion-bonding suffers from the limitation of being unable to readily join systems with stable oxide layers and also requires high pressures, and often long processing times, resulting in high installation and operation costs [3, 4]. On the other hand, brazing possesses the limitation of the remelting temperature of the bond being less than the operating temperature of the superalloys and also of the formation of intermetallics along the bondline [5].

Transient liquid phase bonding is the preferred joining method for nickel-base superalloys because of its better joint properties, lower pressures required for bonding, tolerance of stable oxide layers on the faying surfaces, avoidance of both hot cracking in the joint and (as an isothermal process) the residual stresses that lead to post-weld heat treatment cracking, plus elimination of brittle secondary phases along the joint. In TLP bonding, a liquid forming interlayer (usually rich in the substrate material, plus a melting point depressant) is selected such that the melting point of the interlayer (initially) is significantly lower than the melting point of the substrates and hence starts out in the same condition as a braze joint [5]. TLP bonding has three distinct different stages. 1. Substrate dissolution, which brings the compositions of the interlayer (usually starting from an eutectic) and the solid into local equilibrium. 2. Isothermal

solidification in which the solute diffuses from the interlayer into the substrates under local equilibrium concentrations and constant bonding temperature resulting in re-solidification of the joint at the bonding temperature. Solid-state homogenization to produce a uniform solute concentration and eliminate the formation of secondary phases at the bondline on cooling. Hence, the microstructure and mechanical properties of TLP bonds can approximate those of the substrate material, unlike brazing.[6-11]

This paper discusses the microstructure-mechanical property relationships of TLP bonds between the single crystal superalloy CMSX-4 and polycrystalline IN 738 and IN 939. The joining process used proprietary wide-gap style composite interlayers Niflex-110 and Niflex-115; a conventional foil interlayer BNi-3 was chosen for comparison. The composite interlayers served two functions. One is that these increased the interfacial area between the liquid and solid phases, so that a more efficient path for diffusion of the solute from liquid to solid is created. These interlayers also reduced the amount of liquid necessary to fill the gap so that less solute needed to be diffused to produce isothermal solidification and solid-state homogenization. The differences in the microstructure of the bonds with three different interlayers are discussed in this paper. The joints were tested using the shear rig design developed by Yan and Wallach [12] to determine the room temperature shear strength of the bonds.

2. EXPERIMENTAL PROCEDURE

TLP bonding was performed in a tube furnace under a vacuum of at least 10^{-2} Pa at 1160 °C. All as-received interlayers were 50 μ m in thickness. All the substrates were machined to the required dimensions using electrical discharge machining (EDM). Substrates were then polished using 1000 grit SiC paper and then ultrasonically cleaned in an acetone bath. The substrates were wrapped with tantalum wire (for fixturing) and the assembly was bonded in a tube furnace. All the bonding trials were performed with the (001) plane of the CMSX-4 as the faying surface joined to the polycrystalline substrates, IN 738 and IN 939. A series of interrupted bonding tests (bonding times ranging from 0 min {i.e., heating to the room temperature and immediately cooling}, 1 min, 2 mins, 4 mins, 10 mins, to 1 hr, 2 hrs, and 4 hrs at 1160 °C) were performed to study the bond microstructural development. "Post-bond heat treatment" (PBHT) was performed on the as-bonded samples to allow the diffusion of aluminum from the substrates to form more γ' on the bondline and to reform the \square' distribution in the bulk, which was solutioned during bonding. Post-bond thermal exposures of up to 1 week (168 hrs) at 1000 °C, were used on the as-bonded samples to determine the microstructure-mechanical property relationships of the bond with time. In this paper, the term "post-bond heat treatment" (PBHT) is used to describe treatments to develop the desired γ/γ' microstructure. The standard heat treatments for CMSX-4, IN 939 and IN 738 differ. Hence the heat treatment used for the bonds was a compromise. In contrast "post bond thermal exposure" refers to studies designed to examine the microstructural stability of the bonds. Microstructural studies were done using scanning electron microscopy (SEM), together with energy dispersive x-ray spectroscopy (EDS) while room-temperature mechanical properties were characterized using Vickers micro hardness and shear testing.

3. RESULTS AND DISCUSSION

Microstructural studies showed that all of the wide-gap style bonds formed an eutectic along the grain boundaries of the Niflex core. The formation of this grain boundary liquid appeared to occur in competition with wetting of the faying surfaces, thus leaving non-bonded regions as shown in fig.2. However, this problem was overcome by increasing the boron content of the Niflex foils, so as to form additional liquid.

Given a sufficient holding time at the bonding temperature, isothermal solidification was completed, followed by solid-state homogenization of the remaining bondline boron. Hence,

suppression of bondline boride formation was achieved with the wide-gap style composite interlayers (in contrast to bonds made with BNi-3 foils which contained numerous borides). However, a variety of second phases, apparently borides/carbides and/or topologically close-packed (TCP) phases, were observed to form in the substrate diffusion zone on the polycrystalline side of both the CMSX-4 – IN738 and CMSX-4 – IN939 bonds, as shown in fig.3(a-c) and fig.3(A-C). However, in the case of bonds involving IN738, the formation of these second phases was suppressed by increasing the bonding time to 4 h at 1160 °C for those bonds using either of the two composite interlayers and cannot be avoided in the bonds using BNi-3 foil as an interlayer as shown in fig.3(d, e, f). In contrast, when bonding IN939, the second phases remained stable for all times examined for all interlayers, including post-bond thermal exposure of up to 1 week at 1,000 °C as shown in fig.3 (A-F).

In all bonds examined, there was relatively little gamma-prime at the bondline, when compared with the bulk substrates. For a given wide-gap interlayer, more gamma-prime was formed at the bondline of CMSX-4 – IN939 when compared to that of CMSX-4 – IN738, as shown in fig.4.

3.2 mechanical properties

Mechanical testing was performed on as-bonded, post-bond heat-treated and post-bond thermally exposed specimens. The average peak stress versus the condition of the substrates is plotted in fig.5. The shear strengths of bonds between CMSX-4 – IN939 are higher than those of CMSX-4 – IN738 bonds. Thus, the extent of bondline gamma-prime formation (relatively high in bonds with IN939, low in bonds with IN738) would appear to be more important for room-temperature properties than the presence of brittle second phases in the polycrystalline substrate (invariably present in IN939, but avoidable with a suitable bonding treatment in IN738). Indeed, bonds prepared with IN738 showed an increased shear strength after suitable post-bond heat-treatment (PBHT), to diffuse Al from the substrates resulting in more γ' at the bondline and to reform the γ' distribution solutioned during bonding. Also, bonds that were thermally exposed to 1000 °C for 1 week showed a decrease in shear strength, which can be attributed to the coarsening of γ on the bondline region.

The shear tested specimens were examined in the SEM to characterize the relationship between the microstructure and mechanical properties of the bond. In all of the bonds, a ductile shear fracture was seen on the bondline which can be attributed to the insufficient γ' formation. However, on all CMSX-4 – IN 939 bonds, secondary cracking was seen on the fracture surfaces in addition to the ductile shear as shown in fig.6(a), which was related to the secondary phases in the diffusion zone of the polycrystalline substrate of the bondline. In addition, no secondary cracking but rather dimpling was seen on the fracture surfaces of CMSX-4 – IN 738 bonds as shown in fig.6 (a).

The hardness profile of the bonds is presented in figure 7 in which the Vickers microhardness of the bond line is similar to that of the IN 939, while the CMSX-4 alloy has a higher microhardness than the bond line and the IN 939.

4. CONCLUSIONS

The lack of bonding with the low boron system in the wide gap style composite interlayers led to the non-bonded regions on the faying surfaces due to the competition between wetting of the faying surfaces and the formation of the eutectic along the grain boundaries was observed. Two factors dominated the room temperature mechanical properties of the wide-gap bonds. The first was the extent of gamma-prime formation at the bondline. Results from shear testing and

fractography of the bonds indicated ductile shear failure at the bondline. This was due to the formation of insufficient gamma-prime within the joint, which left a relatively soft bondline region, which was also evident from the hardness data of the bond. The second factor was the presence of second phases in the diffusion zone of the polycrystalline substrate. This led to the formation of brittle secondary cracks. Overall, it is evident that the room temperatures shear strength of the bonds was more dependent on the formation of γ' on the bond line than on the secondary phases in the diffusion zone of the polycrystalline substrate.

REFERENCES

1. S.J.Lee, S.K.Wallace and R.Y.Lin: Acta Mater. 1998, 46, p. 1283.
2. J.R.thamburaj, W.Wallace and J.A.Goldak: Int. Metall. Rev. 1983, 28(1), p. 1-22.
3. A.E.Dray, Ph.D. thesis. 1985, University of Cambridge: Cambridge, UK.
4. X.Wu, R.S. Chandel, and H. Li, J. Mater. Sci. 2001, 36(6), p. 1539-1546.
5. W.F.Gale, and D.A. Butts, Sci.Technol. Weld. Join. 2003, In press.
6. Y.Zhou, W.F.Gale, and T.H.North, Int. Mater. Rev. 1995, 40(5) p. 181-196.
7. T. I. Khan and E. R.Wallach: J. Mater. Sci. 1995, 30, p. 5151-5160.
8. Y.Nakao and K.Shinozaki: Matl. Sci. Technol. 1995, 11, p. 304.
9. M.Tuah-poku, and T.B. Massalski: Metall. Transactions A. 1988, 19A(3), p. 675-686.
10. D S Duvall, W.A.Owczarski, and D F Paulonis: Weld. J. 1974, 53(4), p. 203-214.
11. P.Yan, E.R.Wallach: Intermetallics. 1993, 1, p. 83-97.

FIGURES

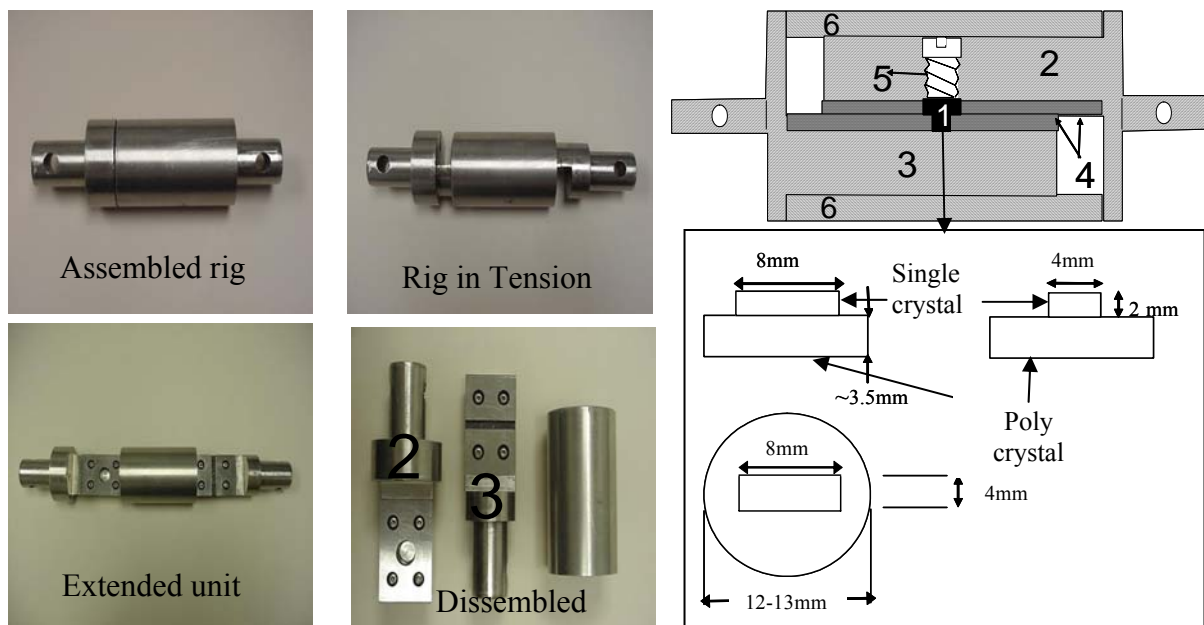


Fig. 1(a)

Fig. 1(b)

Fig.1 (a) shows photographs of the shear rig, with a specimen geometry shown in fig.1 (b). The screw in the fig.1 (b) can be adjusted so that the bond line is placed in the plane of shear and hence mode II shear was applied to the bonds. [13]

1. Sample 2. Down Fixture (circular groove) 3. Up Fixture (Rectangular slot)
4. Tool Steel Hardened inserts 5. Adjustable Screw 6. Shield

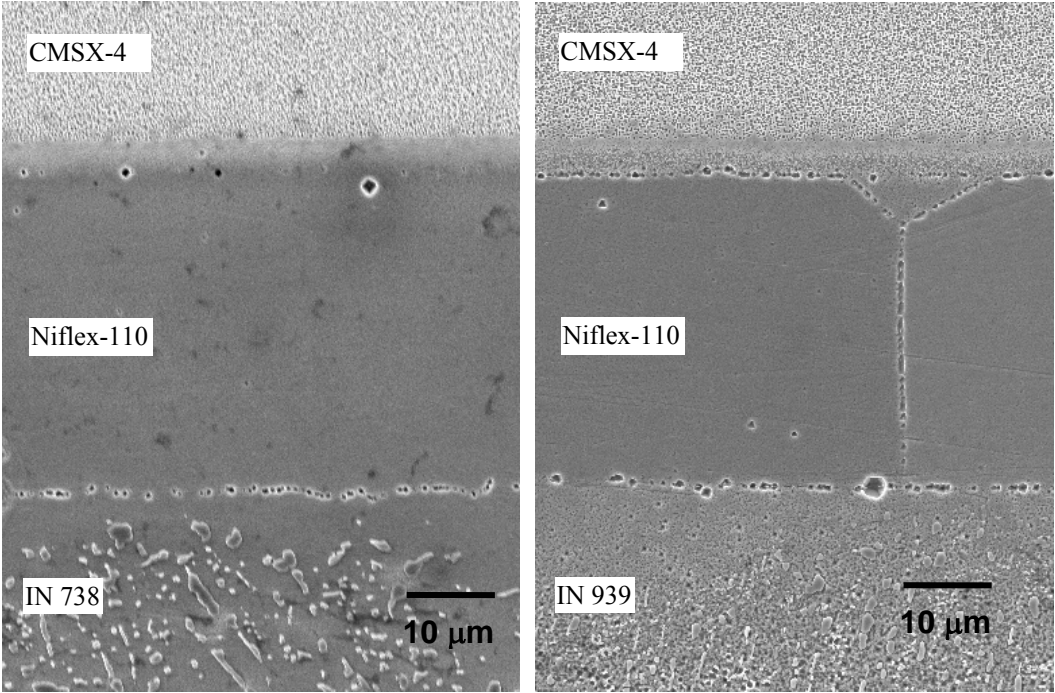


Fig.2 (a).

Fig.2 (b)

Fig.2 (a) CMSX-4 – IN 738 bonds using Niflex-110 after 4 mins of bonding time
Fig.2 (b) CMSX-4 – IN 939 bonds using Niflex-110 after 4 mins of bonding time

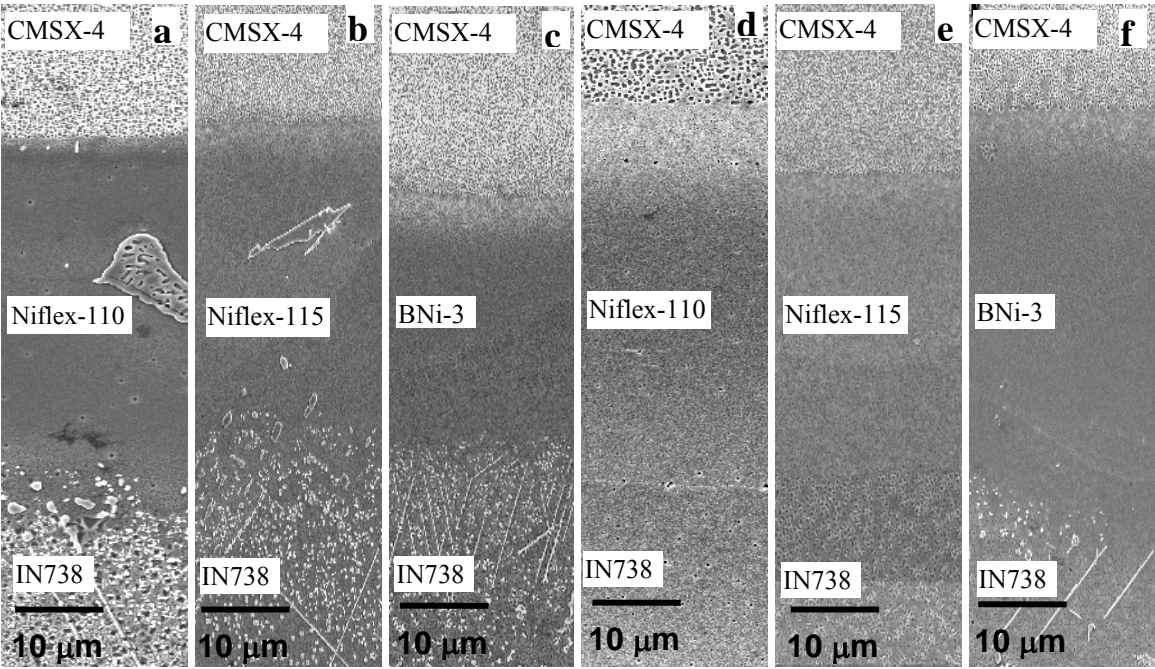


Figure 3

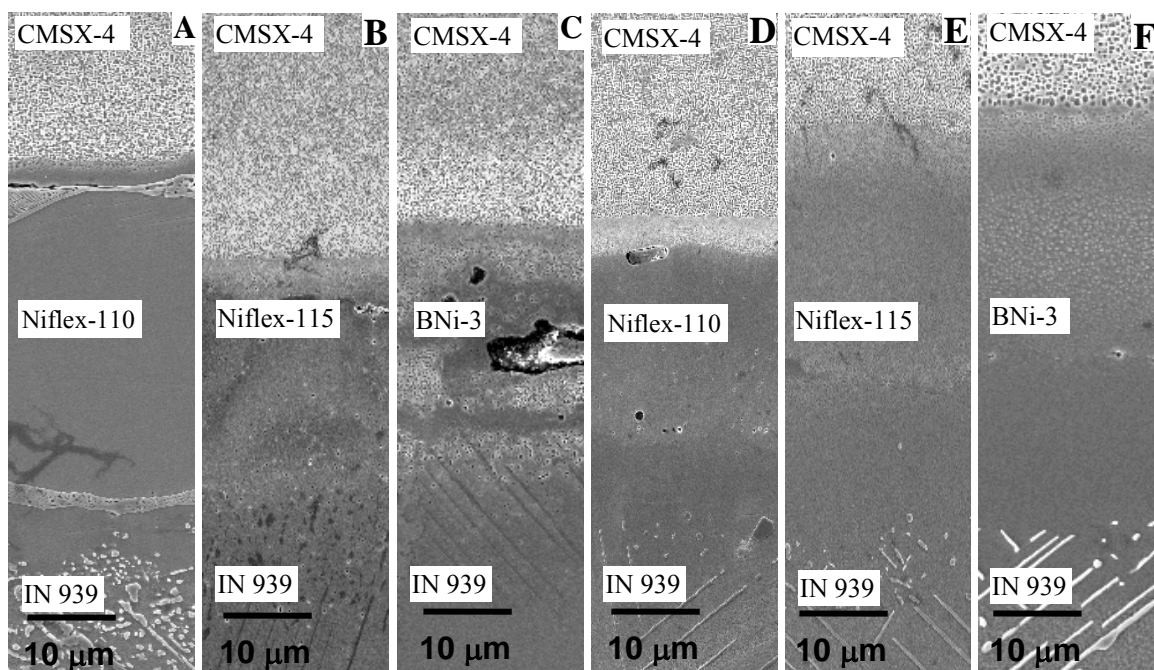


Figure3. Microstructure of bonds: a, b and c represent CMSX-4 – IN 738 bonds using Niflex-110, Niflex-115, and BNi-3 respectively, at 0 minutes of bonding time at 1160⁰C (i.e., heated to 1160⁰C and immediately cooled); d, e and f represent the same combinations for 240 mins of bonding time at 1160⁰C.

A, B and C represent CMSX-4 – IN 939 bonds using Niflex-110, Niflex-115, BNi-3 respectively, at 0 minutes of bonding time at 1160⁰C; D, E and F represent the same combinations for 240 minutes of bonding time at 1160⁰C.

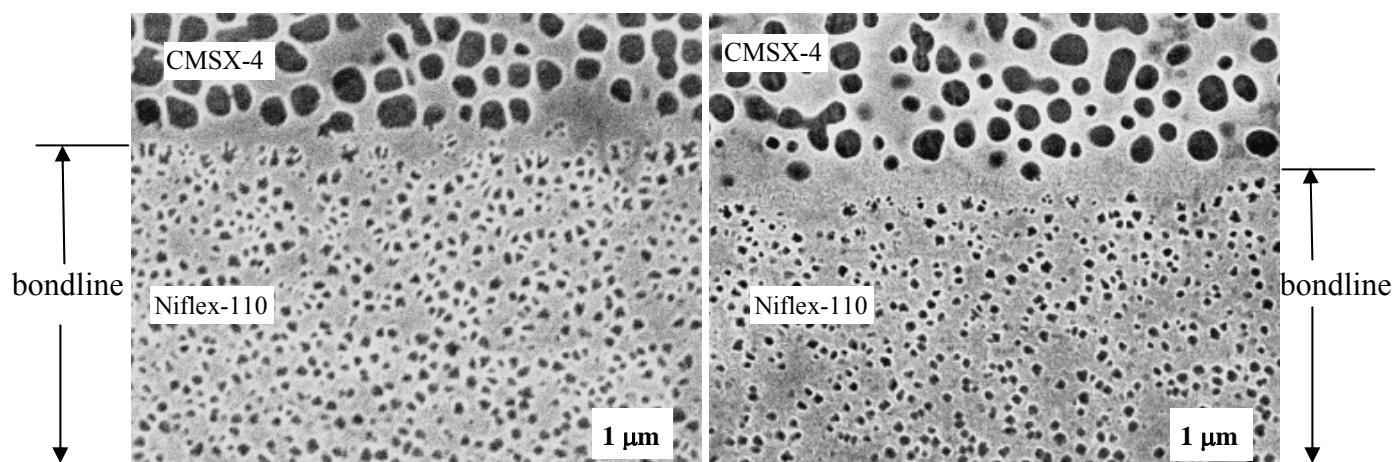


Fig.4(a)

Fig.4(b)

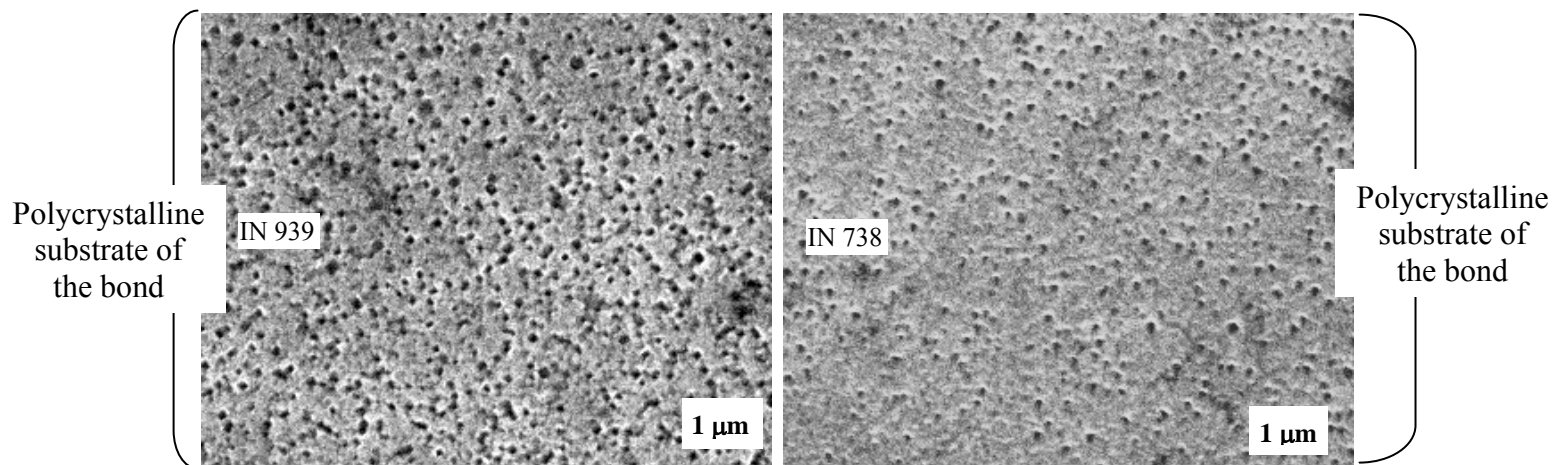


Fig.4(a)

Fig.4(b)

Fig.4 (a).CMSX-4 – IN 939 bond using Niflex-110 at 1160 °C for 240 mins

Fig.4 (b) CMSX-4 – IN 738 bond using Niflex-110 at 1160 °C for 240 mins

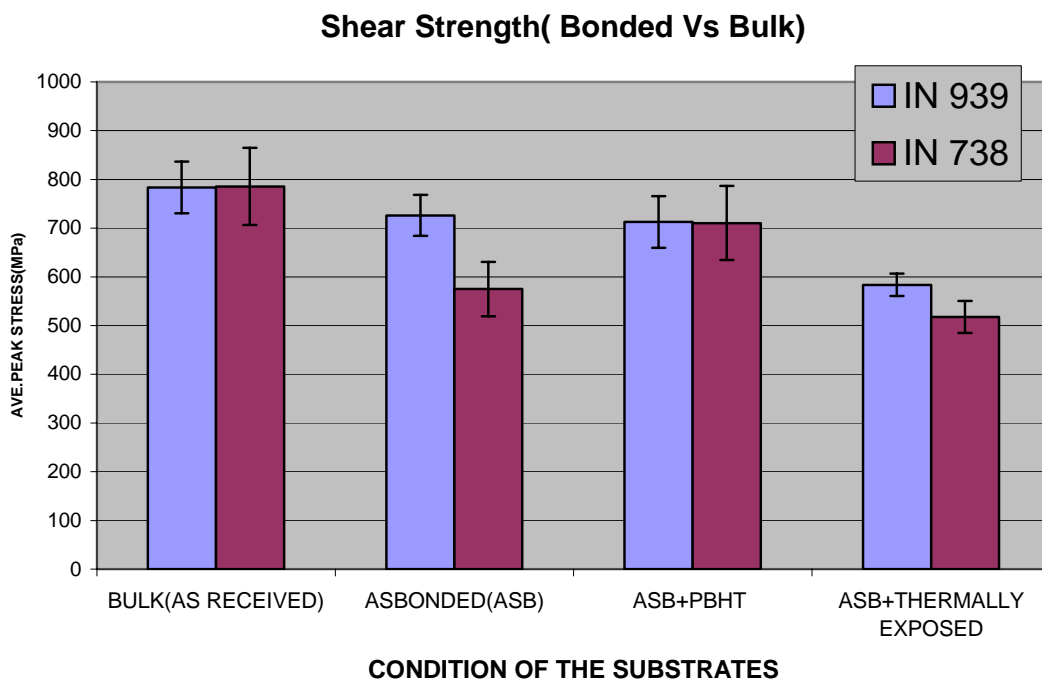


Fig.5 Average shear stress versus condition of the substrates (error bars denote standard deviation)

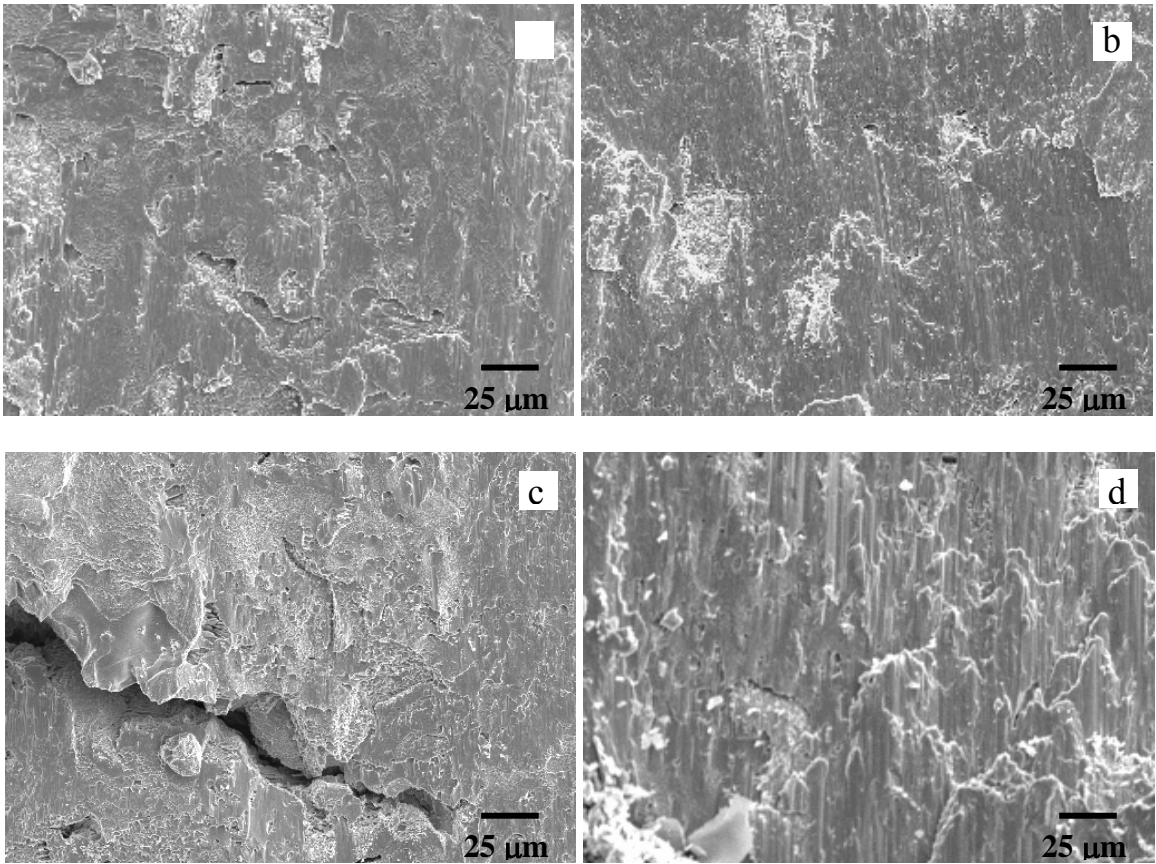


Fig.6 Fractographs of bonds- a, b represent IN 738 and CMSX-4 fracture surfaces of CMSX-4 – IN 738 bonds respectively. c, d represent IN 939 and CMSX-4 fracture surfaces of CMSX-4 – IN 939 bonds respectively.

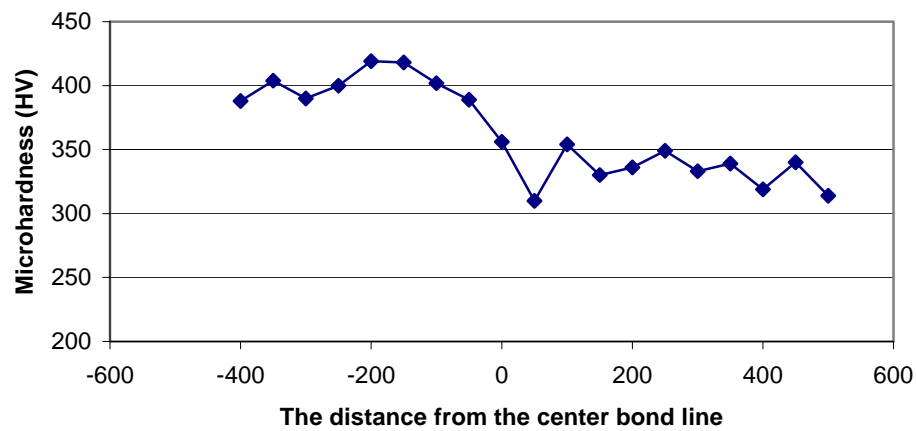


Fig 7: Vickers microhardness across the bond line (CMSX-4 is negative distance and IN939 is positive).