Properties and Fracture Analysis of Diffusion Bonded TiC-Particulate Reinforced Ti-6Al-4V Titanium Matrix Composite

Antonio A. M. da Silva [1], Telmo R. Strohaecker [2] and Jorge F. dos Santos [1]

[1] Joining Technology Group – Institute of Materials, GKSS Research Centre, Max-Planck-Straße, 21502 Geesthacht, Germany
[2] Post Graduation Program in Metallurgy, Mining and Materials/PPGEM, Av. Oswaldo Aranha 99/610, 90035-190 Porto Alegre/RS, Brazil

ABSTRACT

Titanium Matrix Composites (TMCs) offer a combination of good mechanical properties as well as high temperature performance that make them attractive candidate materials for advanced engine components and high temperature structural applications. Joining of such materials is of great concern since the production of complex structures depends on a proven joining technology that produces high quality defect-free joints. Solid state diffusion bonding seems to be an alternative for fusion welding processes since it avoids problems such as segregation effects and undesirable chemical reactions. A metallurgical characterization and mechanical assessment has been performed on Ti-6Al-4V+10%TiC diffusion bonded joints. Fractographic analysis has been conducted on the microflat tensile fracture surfaces in order to investigate the failure mechanism of such materials. Best results were associated with a bonding temperature of 1000ºC, pressure of 5 MPa and 35 minutes of bonding time.

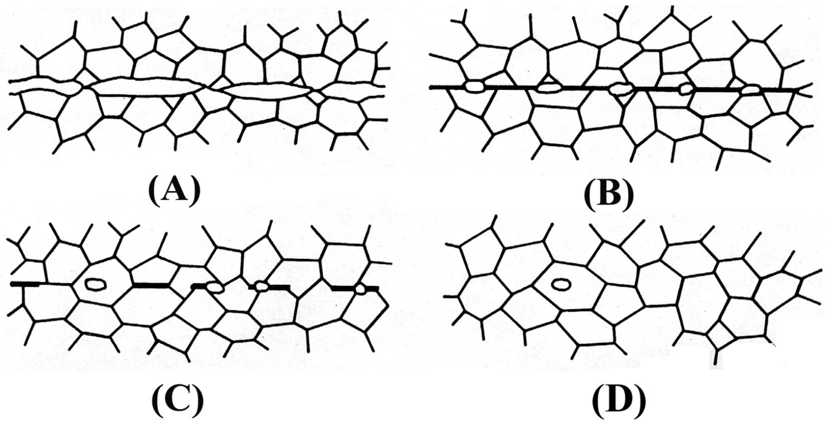
1. INTRODUCTION

Particulate-strengthened TMCs have attracted much attention during the last few years due to its low cost and avoidance of existing problems in the continuously reinforced TMCs. Greatest improvements in mechanical properties are generally obtained using fibre reinforced materials, however, such composites have the disadvantage of being anisotropic, very expensive to manufacture and difficult to manipulate by subsequent forming operations. Particulate reinforced materials tend to offer more modest changes in properties, but are more isotropic and can be processed conventionally. Addition of high hardness and elastic modulus ceramic reinforcements, such as TiC, to low density and corrosion resistant metal matrices, such as Ti-6Al-4V, increases markedly their specific elastic modulus and strength as well as creep-resistant and wear-resistant properties however there is a drastic loss in fracture toughness and ductility [1-6].

Although such materials (e.g. Ti-6Al-4V+10%TiC) can clearly perform even better than the matrix (Ti-6Al-4V) alloy itself, their successful application depends on the availability of proven joining techniques that can produce high-quality joints. Fusion welding processes, such as laser beam welding, have been unsuccessfully applied to TMCs, since there was carbon segregation to the matrix causing joint strength deterioration [7]. Therefore, solid state welding processes such as diffusion bonding would be a natural choice while joining TMCs meeting the requirements for the most critical applications and complex structures. Diffusion bonding process can produce defect-free joints retaining original mechanical properties, such as yield and tensile strength, at room and elevated temperature without any adverse metallurgical reaction [1,2,9-12].

New manufacturing technologies such as the CHIP process (Cold and Hot Isostatic Pressing) can now produce uniformly distributed particle-strengthened TMCs. This innovative process resembles normal powder processing methods. One significant innovation is the use of unalloyed commercially pure (CP) titanium powder and master alloy addition powder which is blended into proper proportions. The last step is a containerless hot isostatic pressing (HIP) at 899ºC, 103 MPa for 2 hours in argon producing full density (typically 99 to 100% density) materials. Another advantage of this process is that the extent of composite stiffening and hardening can be tailored with the TiC content [7,13].

Diffusion bonding is a solid state welding process which produces a weld by the application of pressure at elevated temperature (below the melting point of the material to be welded, usually above 50% of the melting point) with no macroscopic deformation or relative motion of the workpieces [14]. Two necessary conditions must be met before a satisfactory diffusion bond can be made: mechanical intimacy of faying surfaces and the disruption and dispersion of interfering surface contaminants to permit metallic bonding. A three-stage mechanistic model adequately describes weld formation, according to Fig. 1. The first stage is deformation and interfacial boundary formation followed by a grain boundary migration and pore elimination. The final stage is diffusion pore elimination [14].



"Fig. 1 – Three-stage mechanistic model of diffusion bonding. (a) Initial asperity contact, (b) interfacial boundary formation, (c) grain boundary migration and pore elimination, and (d) diffusion pore elimination."

To optimize the structural behavior of structures in composite materials it is necessary to understand the actuating failure mechanism and its relation with the microstructure. According to Mussert [15] the failure mechanism of particle-strengthened TMCs is strongly influenced by the presence of the brittle ceramic phase. It is assumed that fracture in particle-strengthened TMCs follows the same sequence as in structural alloys, namely nucleation at second phase particles followed by failure in the matrix by void coalescence [16]. It is believed that the initial step of void nucleation occurs in the matrix as well, but as a consequence of the increased levels of stress and plastic constraint caused by the addition of the reinforcing particles [15].
The objective of this work is to investigate the feasibility of producing defect-free diffusion bonds in Ti-6Al-4V+10%TiC using four different conditions and to identify the best parameter set based on mechanical and metallurgical evaluation as well as to perform a fracture analysis of the microflat tensile fracture surfaces.

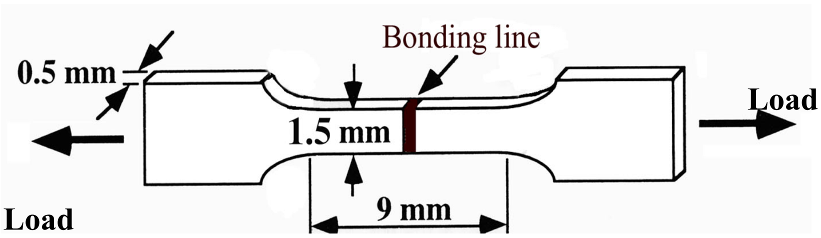
2. EXPERIMENTAL PROCEDURE

The material used in this investigation is the titanium $\alpha+\beta$ alloy Ti-6Al-4V reinforced with 10 wt.% of TiC particles produced by the CHIP (Cold and Hot Isostatic Pressing) process supplied in the form of 50.8 mm diameter bars.
A series of diffusion bonds with different bonding parameters were made using round specimens of approximate size of 50x25 mm^2. The parameters were selcted based on previous experience in diffusion bonding of comparable materials [11]. Bonding temperatures (within the maximum range of the equipment) and bonding pressures ranging from 5 to 7 MPa were used in this investigation (see Table 1). The diffusion bonding surfaces were machined and ground with 1200 grid emery paper and then polished with a Al_2O_3 0.05μm solution followed

by rinsing with acetone prior to bonding. The vacuum was always better than $3.0x10^{-4}$ torr ($4.0x10^{-2}$ Pa) for all bonds.
Scanning Electron Microscopy (SEM) was used to evaluate the microstructural features after diffusion bonding process and to assess bond quality. Microflat tensile specimens (28x1.5x0.5 mm^3 with 9 mm gauge length – see Fig. 2) were cut out from the bonds using spark erosion cutting (Electric Discharge Method – EDM). Tests were conducted in a screw driven Instron 1195 testing machine at 0.1 mm/min cross head speed and the displacement was recorded using a laser extensometer (FOEPS 50 mm). Microflat tensile fracture surfaces were analysed using a SEM in oreder to determine base material and diffusion bonding conditions failure mechanisms.

"**Table 1** – Diffusion bonding parameters used in this investigation."

Condition	Temperature (°C)	Pressure (MPa)	Time (min.)
DB1	875	5	60
DB2	900	6	35
DB3	900	7	60
DB4	1000	5	35

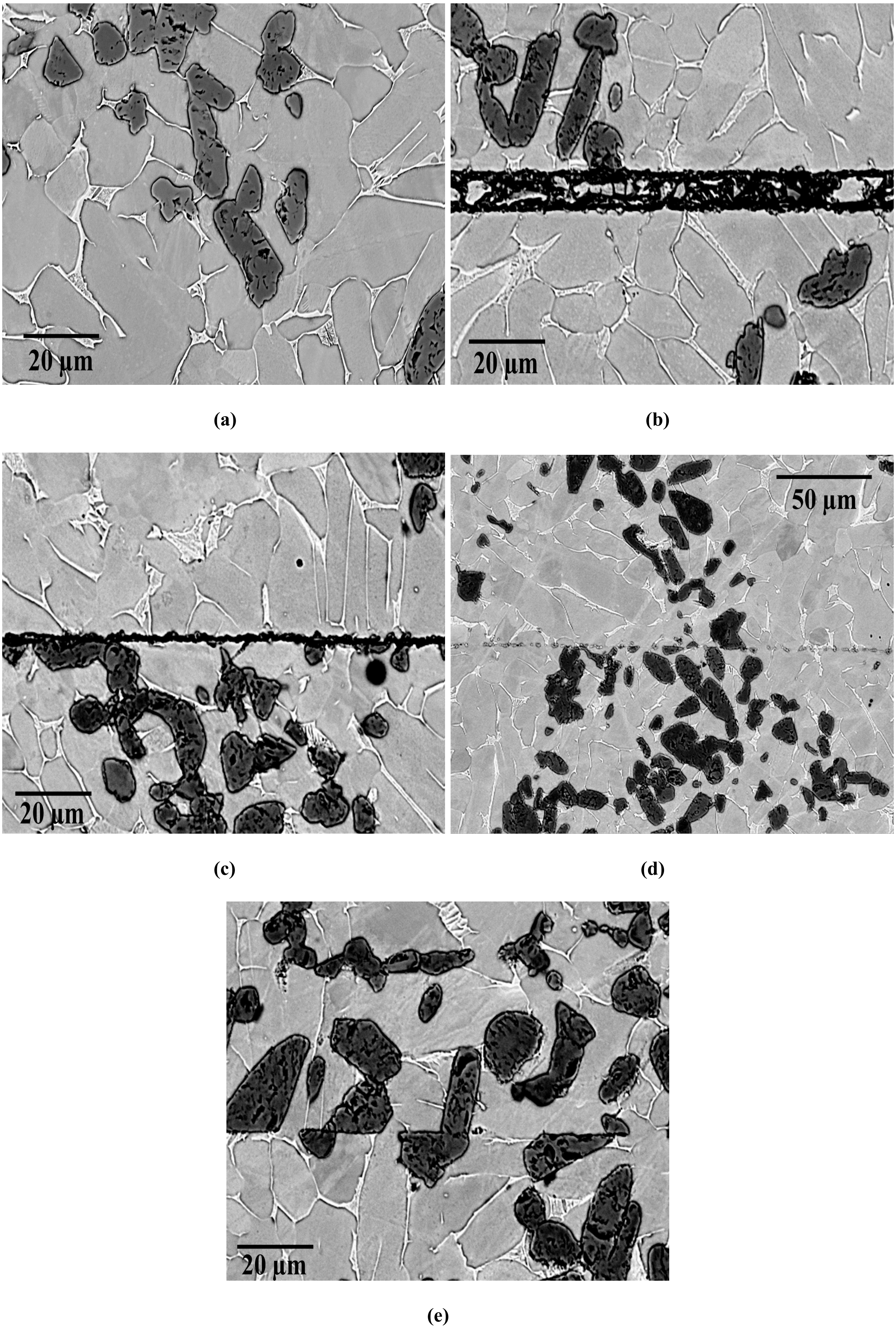


"**Fig. 2** – Microflat tensile specimen geometry."

3. RESULTS AND DISCUSSION

3.1 Metallurgical Characterization

The microstructure of the composite base material and diffusion bonding joints is presented in Fig. 3. It can be seen from Fig. 3a that particulate reinforced base material microstructure consists of platelike alpha (dark grey phase) and intergranular beta (bright grey phase) with TiC particulate randomly distributed in the matrix. The average particle diameter is approximately 15μm and the shape factor is 0.685. Fig. 3b presents bonds made at 875ºC, 5 MPa and 60 minutes and it can be observed that the original interface is clearly visible after diffusion bonding showing the gap between the mating surfaces (unbonded areas). Temperature, pressure and/or time were not enough to allow sufficient diffusion across the bond interface of the mating surfaces. Condition DB2 also presented a visible interface and a gap between the mating surfaces (unbonded areas) showing that even increasing temperature and pressure was not enough to achieve a sound bond, Fig. 3c. It is important to emphasize that in theses two conditions bonding was not fully achieved. Condition DB3 showed a visible bond line in some areas (presence of some remaining voids) but no unbonded areas showing that an increase in the three key parameters (when comparing to DB1) was beneficial to the diffusion bonding process, Fig. 3d. In contrast with the other conditions, DB4 displayed the best results presenting no unbonded areas. The bond interface was indiscernible showing that temperature is an important parameter concerning diffusion bonding, Fig. 3e. All specimens have been prepared using the same procedure before bonding therefore there is no influence of surface roughness in the resulting microstructure.



"**Fig. 3** – SEM back-scattered electron micrographs showing base material and diffusion bonding conditions. (a) platelike alpha (α) and intergranular beta (β) with TiC particles. (b) DB1. (c) DB2. (d) DB3 showing remaining voids and (e) DB4. Kroll´s etchant."

3.2 Mechanical Characterization

Under all bonding conditions investigated no drastic microstructural changes have occurred and composite microstructure remained the same (platelike alpha and intergranular beta with TiC particles randomly distributed) without grain growth or dynamic recrystallization. The only clearly visible change was the presence of few remanescent voids (conditions DB3 and DB4) and also the visualization of the original interface (gap between the mating surfaces) in conditions DB1 and DB2 due to inefficient bonding parameters.

Tensile properties were determined by means of microflat tensile testing in conditions DB3 and DB4 since conditions DB1 and DB2 have not achieved fully bonding. Table 2 presents the tensile behaviour of DB joints as well as base material. As expected DB 4 specimens displayed the best results since no microstructural change had occurred. DB 3 had the worst results since there were some remaining voids at the diffusion bonding line.

All the specimens have fractured without yielding since the presence of TiC particles drastically reduced composite ductility due to the elastic restraint imposed on the matrix by the particulate and the subsequent stress concentration around the particulate (constraint imposed on plastic flow by the particles) [6]. The presence of clustering observed in some areas of the base material composite has also contributed for such behaviour. Table 2 also presents the joint efficiency in terms of ultimate tensile strength ($UTS_{DB\ condition}/UTS_{base\ material}$). It can be clearly seen that condition DB4 presented almost the same UTS values of the base material (99% of joint efficiency) showing the efficiency of the chosen diffusion bonding parameters

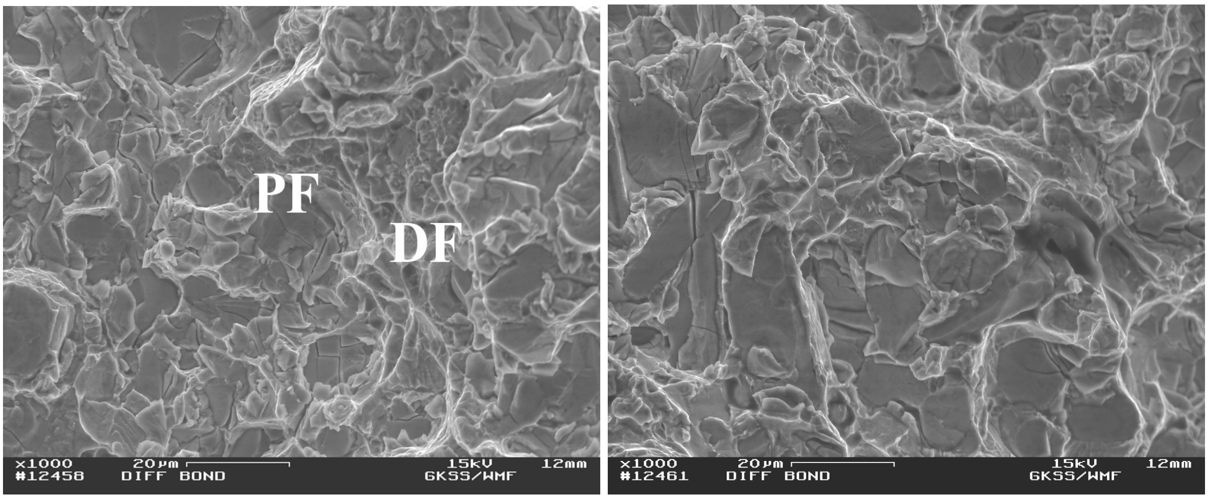
"**Table 2** – Ultimate tensile strength behaviour and Joint efficiency in terms of ultimate tensile strength for the diffusion bonding conditions investigated."

Condition	UTS (MPa)	Joint Efficiency (%)
Base material	846	*
DB3	654	77,30
DB4	838	99,05

3.3 Fractography

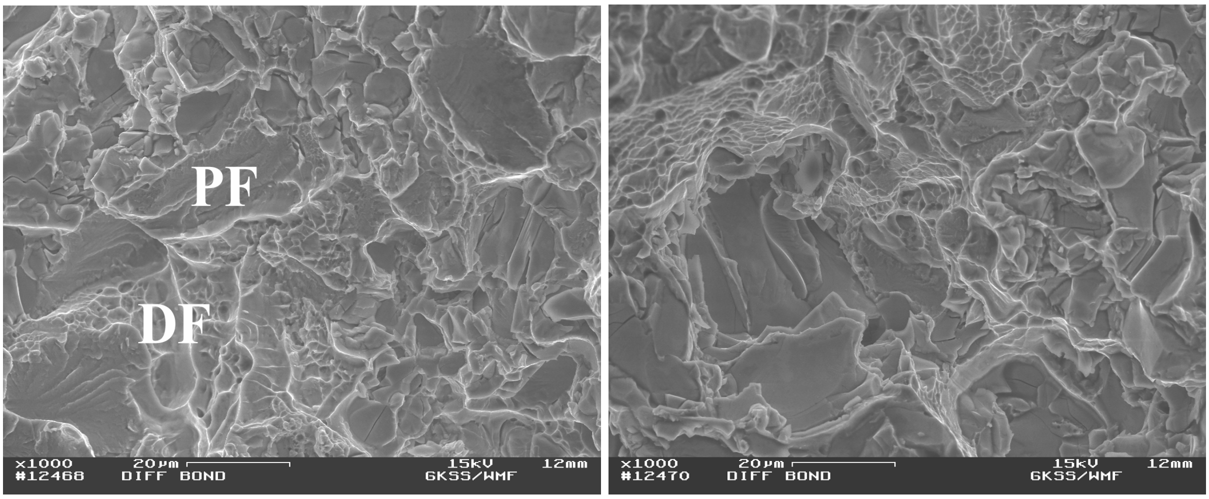
The presence of a brittle ceramic phase in a ductile titanium matrix has effects on the actual failure mechanism. It is assumed that fracture in discontinuously reinforced MMCs follows the same sequence as in structural alloys, namely nucleation at second phase particles followed by failure in the matrix by void coalescence. Such behaviour has been found for the base material and DB4. Another interesting feature is that the initial step of fracture (nucleation) occurs by void nucleation and preferential voids nucleation sites are near or at the reinforcing particle as a result of fracture of the reinforcing particle itself. Void nucleation has also been observed in regions of high volume fraction (clustering) since the elastic reinforcing particles constrain the plastic flow of the matrix, producing increased stress levels in the reinforcing particle and high stress triaxiality in the matrix between the particles. When cracked particles act as void nucleation microcracks are present and these microcracks can be linked to each other causing macroscopic failure [15,17-20]. This fracture mechanism is explained by Manoharan and Lewandowski [21] which have proposed that when the composite is loaded, stress is transferred to the particles. When the stress on a particle becomes large enough it will become damaged (particle fracture). As the applied load is increased more particles reach this stress (in which they become damaged) accumulating internal damage. Such behaviour creates the ideal condition for the appearance of internal defects (incipient voids). The other effect is that the damaged particle no longer supports the load. Thus, as particles become damaged, the stress on the undamaged material increases,

thereby increasing the rate at which the stress on the remaining particles increases. These damaged regions introduce voids which eventually grow and lead to failure.
Although the composite exhibits very limited ductility on a macroscopic scale, SEM fractographies revealed the presence of ductile fracture in the matrix. Since the void initiation mechanism is particle fracture, the crack path followed is the one that contains the highest area density of broken particles, therefore a relatively large percentage of broken particles can be seen in the fracture surface. Figs. 4 and 5 present the fracture surfaces of condition DB4 and base material. It is also possible to observe in the fracture surface the presence of secondary cracking in the particles confirming the mechanism suggested above (internal damage accumulation at or near the particle).



(a) **(b)**

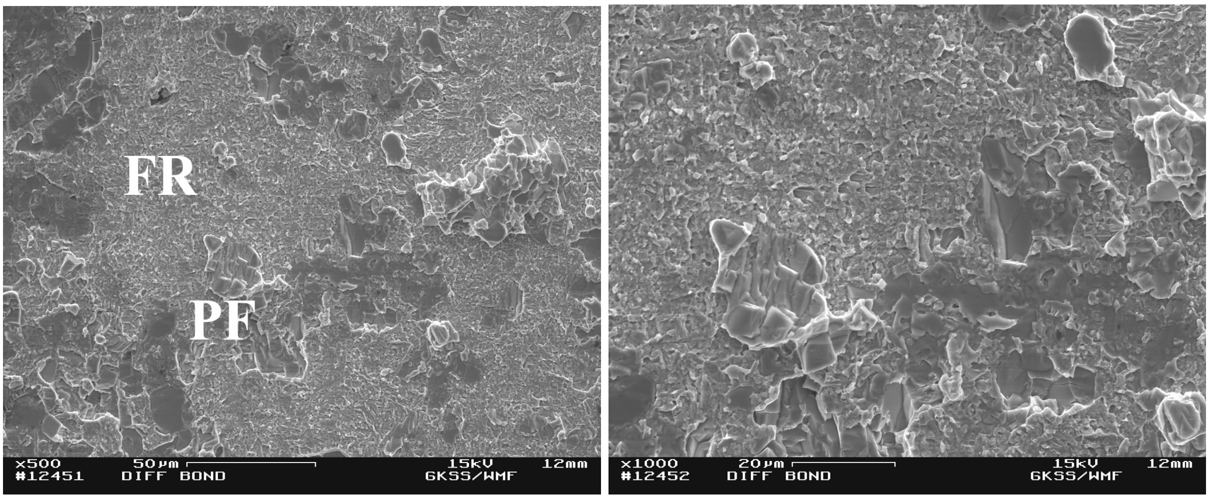
"Fig. 4 – SEM Fracture surfaces from the base material showing locally ductile fracture (dimples) – DF - and particle fracture - PF."



(a) **(b)**

"Fig. 5 – SEM Fracture surfaces from the diffusion bonding condition DB4 showing locally ductile fracture (dimples) - DF - and particle fracture - PF."

Condition DB3 has presented different fracture behaviour from DB4 and base material. Since remaining voids were already present in the specimen (from inefficient diffusion bonding parameters), these irregularities represent a preferable crack path contributing for the reduction of the tensile strength and helping the accelaration of the damage process. It can be seen from Fig. 6 that flat and some rough regions are present at the fracture surface of the microflat tensile specimens. The rough regions correspond to the fracture in the composite (particle fracture and brittle fracture); whereas, the flat regions, which predominate the fracture surface, correspond to the fracture along the bond line.



(a) **(b)**

"Fig. 6 – SEM Fracture surfaces from the diffusion bonding condition DB3 showing flat regions (fracture in the bond line – remaining voids) - FR - and rough regions (particle fracture and brittle fracture) - PF."

4. CONCLUSIONS

The effects of different bonding conditions in a Ti-6Al-4V+10%TiV alloy have been analyzed by means of microflat tensile testing and metallurgical characterization. Differences have been found in the fracture behaviour of two diffusion bonding conditions and base material. The following conclusions summarizes our findings:

(1) No microstructural change or transformation has been found due to the diffusion bonding precess when appropriate parameters have been used The "after-weld" microstructure remained the same as before welding consisting of platelike alpha, intergranular beta and TiC particles randomly distributed in the matrix.

(2) Defect-free bonds could be achieved using bonding conditions of 1000°C, 5MPa with bonding time of 35 minutes (condition DB4).

(3) Best tensile properties were displayed by condition DB4.

(4) Particle cracking is the dominant damage process prior to final fracture. Fracture of the particles coupled with the failure of the matrix is the dominant fracture mechanism of the base material and best diffusion bonding condition.

ACKNOWLEDGEMENTS

The authors of this paper would like to thank Mr. K.-H. Bohm for his help in conducting the diffusion bonding process.

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