FATIGUE BEHAVIOUR OF AN ALUMINIUM ALLOY MATRIX COMPOSITE

ANA GARCIA ROMERO*, M. ANGLADA** AND A. M. IRISARRI*
*F. INASMET, Mikeletegi p.2 20009 San Sebastian (Spain) airisar@inasmet.es
**ETSEIB-UPC, 08028 Barcelona (Spain)

Abstract

The fatigue behaviour of an aluminium alloy matrix composite locally reinforced with 15% short δ-alumina fibres has been analysed. Prototype pistons were industrially produced by squeeze casting infiltration of a preform of fibres with the molten Al-12%Si-1%Mg-1%Cu-1%Ni alloy into a cylinder shaped mould. Samples were in the T5 condition that is just aged for 12 hours at 160º C. Fatigue tests were performed with a stress ratio R=0.1 leaded to an average value of the Paris exponent of 9.5 with a relatively reduced scattering, similar to that found in the fracture toughness tests. This value of the exponent is higher than those usually found in aluminium alloys but is low for a brittle composite. Analysis of the data revealed that crack closure played a significant role in the fatigue crack growth in this condition. The use of scanning electron microscopy helps to find an explanation to this behaviour.

Introduction

Metal matrix composites are complex materials constituted by a certain volume fraction of continuous or discontinuous ceramic reinforcement fibres inserted in a metallic alloy matrix. The load transfer between matrix and reinforcement fibres, which allows to obtain advantage of both the high strength and the stiffness of the fibres and the ductility of the metallic matrix, characterizes the mechanical behaviour of these composite materials. Locally reinforced components are very attractive at an industrial scale because the characteristics offered by the composite are employed only in the places where they are actually required, without suffering the drawbacks of lower toughness, higher heterogeneity, and higher cost that are usually found when the reinforcement fibres are present in the whole component.

It is well known that maximum strength of aluminium alloys containing silicon and magnesium in their composition is achieved in the solution treated and aged to the maximum strength condition designed as T6. This treatment leads to the presence of fine and uniform distribution of Mg_2Si particles that strengthen the alloy. In the case of δ-alumina short fibres reinforced aluminium alloys similar results have been reported, showing no relevant influence of the Mg_2Si depletion and silicon enrichment during the T6 heat treatment [1-3]. It has been pointed out that, generally, the heat treatments of the discontinuously reinforced aluminium alloy composites should be performed at the same temperatures than those used for the unreinforced alloys, but with shorter holding times due to the large microstructure distortions and the residual stresses present in the composite accelerate both the solution and the ageing processes [3].

Nevertheless, it must be remarked that all these studies have been performed on laboratory produced composites, using small preforms and low-alloyed materials, together with laboratory, more controlled, production conditions. However, when industrial components must be produced the selected alloys are usually much more complex. As result the number of reactions between the various alloy elements and impurities present in these materials and the preforms that could take place is significantly increased. Additionally, the preforms required in these industrial components are larger than those employed in laboratory works and higher the binder content which is required to keep the ceramic fibres together, without deformation under the metal infiltration conditions. As a consequence of all these difficulties it was completely impossible to obtain a sound material after T6 treatment and the T5 condition, where the solution treating is not applied and the risk of reaction between magnesium and the silica present in the fibres as a binder prevented, was chosen [4].

Aluminium alloys reinforced with particles or whiskers of SiC generally exhibit better near-threshold fatigue crack growth characterisitics than the unreinforced matrix alloy if fatigue failure occurs predominantly within the ductile matrix [1]. Deflection of the fatigue crack by the brittle particles, enhanced crack closure and crack trapping contribute to this improved fatigue resistance in the composite by lowering the effective crack-tip opening displacement. However, if the particular combination of processing conditions, particle size and ageing treatment promotes particle fracture or interfacial separation, the fatigue crack growth resistance of the composite can be significantly lower than that of the matrix alloy [5].

A large number of research works have been devoted to the study of fatigue on aluminium alloy composites reinforced with particles or whiskers. Nevertheless a very reduced of studies have been dedicated to the fatigue behaviour of aluminium alloy composites reinforced with short fibres [6] and no one on industrial highly alloyed matrix composites.

Experimental procedure

Materials

The prototype pistons were industrially produced by squeeze casting, infiltrating a so-called preform of fibres with the molten aluminium alloy into a cylinder shaped mould by Sidenor Reinosa (Spain) according to their proprietary process. The disc shaped preforms with 200mm diameter and 20mm thickness used for this study were supplied by Vernaware (Switzerland) and contained nominally 15% in volume of δ-alumina short fibres, binded by colloidal silica. Chemical analysis revealed a 15% weight of SiO_2 in their composition. The aluminium alloy used in this work conforms to the AFNOR AS12UNG type, widely known as the “piston alloy”, and was supplied by Inespal (presently Alcoa, Spain). Its chemical composition is shown in the table 1.

TABLE 1: Chemical composition of the alloy employed in the prototype pistons

Percentage of alloying elements										
Si	Cu	Mg	Ni	Ca	Fe	Mn	Pb	Sn	Ti	Zn
12,05	1,24	0,98	1,05	0,002	0,36	0,04	0,002	0,001	0,008	0,009
Balance Al										

The preforms were preheated at 600ºC, located into the mold, and infiltrated with the molten alloy at 590ºC under direct axial pressure. Accelerated cooling was supplied to increase the solidification rate. Due to the production route which basically consisted in pouring a dispersion of fibres in water suspension into a filtering mould, where they formed a rigid porous body of approximately 15% volume fraction of fibres and 85% of air, the fibres were distributed in the so called random-planar orientation. This means that the fibres were mainly located in layers, thus producing their preferential location in the stacking plane x-y [7]

Heat treatment

The heat treatment intended to be used for these piston was the T6 consisting in solution treating and ageing, that yields the maximum strength. Nevertheless, due to the presence of clear signs of micromelting in the composite the T5 condition and the influence of both temperature and time of ageing on the mechanical properties investigated [8]. This study pointed that material aged for 12 hours at 160º C gives the optimum properties and this treatment was chosen.

Mechanical tests

Cylindrical 6.4 mm diameter and 25.4 mm gauge length smooth specimens were machined from the composite zone of the pistons. Two of these specimens were used to perform tensile tests at room temperature according to ASTM D3552 standard. The remaining ones were used for conducting uniaxial constant stress amplitude fatigue tests according to ASTM E466 standard. These fatigue tests were carried out also at room temperature with a stress ratio R=0.1 and a frequency of 10 Hz. The stress-life (S-N) diagram was obtained by plotting the maximum stress of the cycle versus the fatigue life.

Moreover, CT 10 mm thick and 20 mm wide specimens were taken to study the fracture and fatigue crack growth behaviour. Chevron notch was chosen to obtain a more homogeneous crack front as marked curvature has been observed when a straight through notch is used [9]. Fracture toughness tests were carried out at temperatures ranging from 25 to 250º C according to ASTM E399 standard.. However all the fatigue tests were performed at room temperature under two different stress ratios R=0.1 and R=0.5, according to ASTM E647. It must be remarked that the geometry of these very small specimens induced some problems. Firstly, the above indicated standard recommends that the ratio between the wideness and the thickness of the specimen was at least 4, that is double than the value used in the present study. Nevertheless, a ratio down to 2 is admitted if the effect of through thickness curvature is corrected. Secondly, and more important the crack can only grow a short length before the accelerated final failure is reached and the number of data available is reduced, invalidating the results of some tests, as the failure is produced before enough data for a real crack growth evaluation are obtained. In any case, that was the only possible geometry as the shape of the prototype pistons preclude the use of larger specimens or chosen another configuration.

Fractographic examination

After failure a scanning electron microscope examination of one half of all these specimens was carried out to determine the failure operating mechanism and the phases that have contributed to the failures were identified by energy dispersive spectrometry. Additionally, metallographic samples were obtained from the other halves, in a perpendicular orientation to the fracture surfaces and including them. These metallographic samples were observed in both optical and scanning electron microscope.

Results and discussion

Table 2 shows the average values of Young Modulus, yield strength, ultimate tensile strength and elongation, obtained in the composite material.

TABLE 2: Room temperature tensile properties of the composite material

Young Modulus (Gpa)	Yield Strength (MPa)	Ultimate Tensile Strength (MPa)	Elongation (%)
107	255	261	0.44

As it was emphasized in a previous work the poor ductility of the base alloy (0.58%) , associated with the presence of large particles of brittle phases was even decreased in the composite material, where both the fibres and the needle like particles of Al_6Cu_3Ni phase, not observed in the base alloy, contributed actively to the fracture [8]. Fractographic examination of the broken specimens confirmed that failure progressed in the composite mainly by the brittle fracture of the ceramic fibres and this intermetallic phase although some debonding of the fibres when they were sited parallel to the fracture plane was also observed.

Totally 72 CT specimens were machined for fracture toughness and fatigue crack growth evaluations but 33 (45% of them) broke during the fatigue pre-cracking, failure that was associated with the brittleness of the composite. However, due to the use of a Chevron notch, all the fracture tests but one yielded valid K_{IC} results. The average room temperature K_{IC} values is low (just 8.7 $MPa.m^{1/2}$) and provides a good explanation for the frequent failure of the specimens during the pre-cracking step. At higher testing temperatures a slight increase(around 12%) in fracture toughness is observed. This value is kept near constant, with only a slight decrease, up to 200º C, but a more marked loss is found when the material is tested at 250º C, falling the values below those recorded at room temperature. A detailed analysis of the effect of the testing temperature on the composite fracture toughness was given in a previous paper [7].

Table 3 summarizes the results obtained in the constant amplitude uniaxial fatigue tests. It must be remarked that even if maximum care was taken in the machining of these specimens, polishing them in the lath to obtain a very good quality surface finish, one failure was originated at a machining notch. This result is not included in the table.

TABLE 3: Stress amplitude versus fatigue life

Stress amplitude (MPa)	No. cycles to failure	Stress amplitude (MPa)	No. cycles to failure
165	50695	146	24775
152	48298	146	98954
151	67534	139	176088
146	698954	144	122014*

* no broken later 30000 cycles at 151 MPa and 13614 cycles at 188 MPa to failure

A first glance to this table evidences the large scatter of the results with a ratio of near 30 between the maximum and minimum lives at the same stress amplitude. Scanning electron microscope examination of the failed specimens revealed a rough surface, with the presence of clusters of fibres or other discontinuities at the origins of the failure. Generally, these failure origins were sited at the surface of the specimens and indicates that the presence of a cluster of fibres at periphery of the component could significantly decrease its fatigue performances. Moreover, some secondary cracks, that were not found in the fracture surfaces of the tensile tests specimens, were also observed.

Table 4 exhibits the Paris law exponents determined in the 11 crack propagation tests performed, 5 with stress ratios R=0.1 and 6 with a stress ratio R = 0.5. It must be indicated that other 11 tests were also carried out but they did not gave valid results due to premature failures, without recording enough data to obtain a crack growth law, of the specimens. This comment evidences the difficulty as only half of the crack growth tests yielded valid results.

TABLE 4: Paris law exponents determined in the crack growth tests.

Specimen	Stress ratio R	Paris exponent	Specimen	Stress ratio R	Paris exponent
63.3	0.1	12.88	65.3	0.5	17.37
63.2	0.1	10.44	60.3	0.5	14.94
59.3	0.1	7.01	64.3	0.5	7.51
79.1	0.1	8.28	64.1	0.5	32.37
66.3	0.1	8.94	76.1	0.5	9.37
-	-	-	78.2	0.5	12.01

These results exhibit a relatively reduced scattering, very similar to that found in the fracture toughness tests and significantly lower than the enormous one observed in the constant stress amplitude uniaxial fatigue tests. The average value of the Paris exponent (9.5) is higher than those usually found in aluminium alloys but it can be considered low for a brittle composite. Analysis of the data revealed that crack closure played a significant role in the fatigue crack propagation in this condition. It is well known that crack closure can have a dominating influence on the fatigue crack growth at low stress ratios, mainly when the propagation follows a tortuous path through the interfaces between reinforcement fibres and matrix.

Crack deflection has been viewed as one of the mechanisms for the toughening of brittle and ductile matrix composites, when the obstacles in the propagation path of the crack, for example the reinforcement fibres, may cause apparently beneficial resistance to crack growth by twisting the crack front [10]. It has been established that even small deflections in the fatigue crack path can lead to a reduction in crack growth rates, especially in the near-threshold fatigue regime [5]

This type of failure was confirmed by the scanning electron microscope observation of the fracture surfaces. In the initial steps of the crack propagation a large number of fibres, that are debonded from the matrix, are found lying on the fracture surfaces. Preferential crack growth along the matrix – fibre interfaces has been reported in a previous work, pointing that this mechanism is specially active in the near-threshold condition [11]. Considering this important contribution of the matrix – fibres interfaces in the fatigue crack propagation the strong influence of the orientation of these interfaces on the fatigue performances is evident. As a consequence of this crack deviation the apparent crack growth rate is lower in the composite than in the base alloy [12].

As the crack growths a change in the failure mechanism with a significantly less marked contribution of the reinforcement fibres to the failure is detected. Moreover, this fractographic analysis revealed the noticeable influence of the large second phase particles, especially those identified as $Al_{23}Ni_5Fe_3Si_2Cu$, on the preferential crack paths. Less alloyed materials do not present these phases in their microstructure and, consequently, the contribution of this kind of particles on their failure has not been reported in other papers. When the crack progression is promoted by the fracture of particles, two contradictory effects are observed; the sudden crack jump when the particles are broken and the retardation associated to the bridging or the roughness induced crack closure [13]. It can be argued that the same particles that are observed in the microstructure of the composite are also in the matrix alloy. Even if this is true the presence of silicon precipitates is higher in the composite while the Mg_2Si particles are finer and lower in number [4].

Additionally, some large defects that are present in the industrially produced components, such as very long fibres, have been found in the fracture surfaces, indicating that they played an active role in the failures, accelerating them. Finally, as it was observed in the smooth specimens tested under uniaxial fatigue stresses, a large number of secondary cracks were present in these fracture surfaces.

Usually there are considered short cracks those whose length is similar to the size of the grain or the metallurgical facet that controls the process. However, in the particles reinforced composites the frontier between short and long cracks is not clearly defined and it can be associated with the change in the crack propagation mechanism [14,15]. Initially, small cracks are nucleated in various zones, generally at the interfaces, at the particles or fibres or at the porosity which grow and coalesce until they reach a certain size when the stress intensity factor promotes their continuous and stable propagation [16].

A clearly higher average value of the Paris exponent and a much larger scatter is observed in those fatigue tests carried out with a stress ratio R=0.5. In this case an average value of 15.6 (more than 64% higher than that obtained in the R = 0.1 tests) was determined. Actually, the analysis of the plots of crack growth versus the stress intensity factor amplitude indicates that no real stable crack growth is produced in most of these tests and consequently the Paris law cannot be used for an accurate evaluation of the fatigue performance of the material in this condition.

Fracture progresses by the debonding between the matrix and the fibres along the interfaces, during the whole failure process, without any noticeable sign of crack closure. Crack progresses along the matrix – fibres interfaces, mainly along those which are sited in the crack growth direction. An apparent contradiction arises from these results as the failure mechanism observed in these specimens, where crack closure is practically inexistent, is nearly identical to that found in those tested under a stress ratio of R = 0.1 at the origin of the failure, when crack closure actively participates in the crack propagation process.

Conclusions

a. The fatigue behaviour of an Al-12%Si-1%Mg-1%Cu-1%Ni alloy matrix composite locally reinforced with 15% short δ-alumina fibres has been analysed. A large scatter in the fatige lives is observed.

b. Fractographic examination indicates that the presence of a cluster of fibres at periphery of the component could significantly decrease its fatigue performances

c. Tests performed with a stress ratio R=0.1 leaded to an average value of the Paris exponent of 9.5. Analysis of the data revealed that crack closure played a significant role in the fatigue crack growth in this condition

d. Crack closure can have a dominating influence on the fatigue crack growth at low stress ratios, mainly when the propagation follows a tortuous path through the interfaces between reinforcement fibres and matrix, mechanism that is confirmed by the fractographic examination at the scanning electron microscope.

e. A clearly higher average value of the Paris exponent and a much larger scatter is observed in those fatigue tests carried out with a stress ratio R=0.5. The analysis of the plots of crack growth versus the stress intensity factor amplitude indicates that no real stable crack growth is produced in most of these tests and consequently the Paris law cannot be used for an accurate evaluation of the fatigue performance of the material in this condition.

f. Fractographic examination of these samples revealed a fracture topography very similar to those observed in the origin of the failure in the samples tested under R=0.5 stress ratio even if the crack closure and crack growth rates are completly different in ones or others. No clear explanation has been found for this similitude and indepth research is needed before a conclusion colud be reached.

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