FRACTURE TOUGHNESS OF AN ALUMINIUM ALLOY MATRIX COMPOSITE

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ABSTRACT

The influence of the temperature on the fracture toughness of an aluminium alloy matrix composite has been studied. The material chosen for this study consisted in a AS12UNG alloy locally reinforced with 15% short δ-alumina fibres, Saffil (trademark of ICI). The material was produced by direct squeeze casting infiltration of fibre preforms (160 mm diameter x 20 mm thick) in a cylinder shaped mold producing 30 mm thick/160 mm diameter discs where thw lower part was the composite material. The discs were heat treated to the T5 condition (artificially aging), and testing specimens were extracted from the composite zone. The fracture toughness characterisation was performedn according to ASTM E399, in the temperature range between 20ºC and 250ºC. CT specimens 20 mm thick were machined with chevron notches, their surface was polished, and they were precracked at room temperature. All the tests performed but one were valid for the determination of K_{IC}. The fracture surfaces and profiles were observed, showing a clear influence of the temperature in the operating fracture mechanism. The stress intensity factor corresponds to a brittle material, but with specific mechanisms due tot eh presence of fiibres.

INTRODUCTION

Metal matrix composites are intrinsically heterogenous materials, and even in the case of particulate composites their behaviour do not usually correspond to that of isotropyc materials. In the case of the present work, there is a specific anisotropy in the z-axis of the material evaluated. This material was produced by squeeze casting infiltration of δ-alumina short fibre (Saffil, trademark from ICI) preforms with the aluminium alloy AS12UNG. The preforms employed were commercially purchased from Vernaware. Due to their production process, which basically consists in pouring a dispersion of fibres in a water suspension into a filtering mold where they form a rigid porous body (of aprox. 15% volume of fibres and 85% air), the fibres in the preforms are distributed in a so called "random-planar" orientation. This means that the fibres are mainly located in layers, thus producing preferential location in the stacking plane x-y. Therefore, the composites produced with these preforms are intrinsically anisotropic in the z direction (direction of infiltration.

The aim of the present work has been the determination of the stress intensity factor KIC, of the composites materials, and its evolution with temperature up to 250ºC. The tested direction corresponds only to the x-y plane due to the impossibility of extracting specimens in the z-x or z-y planes. However, several considerations have to be taken into account when talking about the stress intensity factor of composite materials, and high caution has always to be taken when employing those data in design considerations:

First of all, the anisotropy of the composites is a large difficulty in the determination of a reliable value of the stress intensity factor, K_{IC}. According to the lineal elastic fracture mechanics theory, the measurement of this factor requires that both the testing parameters and the specimens fulfill Mode I plane strain conditions, neglecting other type of phenomena, like the crack tip plasticity, in the beginning of instable crack propagation. K_{IC} is usually determined following the conditions of the ASTM E399 standard, which is prepared according to the intrinsic properties of metallic materials. It is therefore assumed that the materials are: homogenous, isotropic, with an elasto-plastic behaviour, which does not correspond to the composite materials. An additional difficulty is usually the small volume of the test specimens, which in the case of high anisotropic materials generally leads to a mixed mode of fracture instead of the intended mode I fracture [1,2].

Finally, to sum up to these dificulties, the stress intensity factor corresponds to a typically brittle crack propagation process, where the energy absorbed in the fracture process is mostly employed in the crack propagation, and therefore this occurs in an instable typically brittle mode. When the materials have other energy absorbing mechanism, like plastic deformation, other theories were developed and other type of parameters are measured instead the stress intyensity factor. This is the case of the integral J, usually employed in high ductility metals. However, this is not case of composite materials, where energy is usually absorbed by mechanisms of fibre pull-out, crack branching and crack deflection, and others.

EXPERIMENTAL DESCRIPTION

Materials

The material evaluated was an experimental composite produced in Sidenor (Spain) according to their production parameters. The composition of this material are shown in table 1. The as cast material was heat treated to the T5 condition (artificial aging, 160ºC, 12h), and the testing specimens were subsequestly machined. The mechanical properties of the composite material in the T5 condition are shown in table 2. The evaluation of the reliability and preproducibility of these characteristics is reported in [3].

TABLE 1
COMPOSITE COMPOSITION

Percentage of component	Composition
15% Saffil Fibres Preforms	85% Saffil fibres of δ-alumina (97% Al_2O_3, 3% SiO_2) 15% SiO_2 coloidal
85% Alloy AS12 UNG	12% Si, 1,24% Cu, 0,98% Mg, 1,05% Ni, 0,36% Fe, bal. Al

TABLE 2
COMPOSITE MECHANICAL PROPERTIES

Composite	U.T.S. (MPa)	Y.S. (MPa)	E (GPa)	Total elongation (%)
Untreated	236,5	207,5	87,2	0,54
T5	261	255	107	0,44

The selection of the T5 heat treatment was due to micromelting problems observed during the standard solution treatment T6. The estudy of the effect of the different treatments has been reported elsewhere [4]. However, the T5 treatment in aluminium alloys is generally employed for homogenisation of precipitates and for the release of the residual stresses. In the present case, the extent of the T5 treatment is larger than usual due to the fast solidification under pressure produced by the squeeze casting process. This leads to a supersaturated solid solution of most second phases, and specially of Silicon and Mg_2Si. The T5 treatment

leads to a precipitation of the second phases in small size precipitates, both in the alloy and in the composite. Two main differences were observed when compared to the unreinforced alloys and the composites: First, an important number of primary silicon precipitates was produced in the composite, while they were scarce in the alloy. Second, the Mg_2Si precipitated mostly surrounding the fibres (figures 1 and 2).

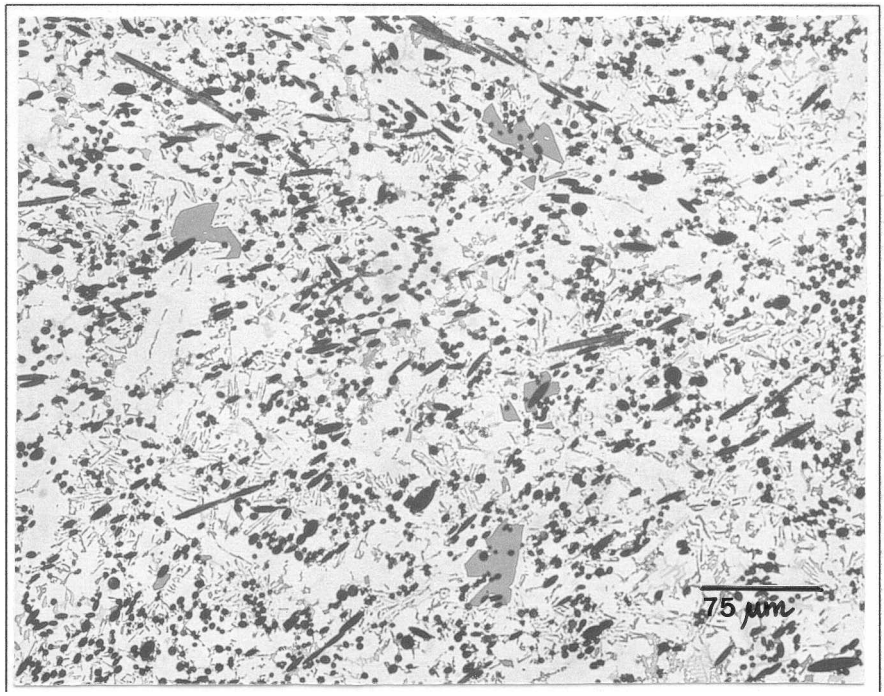

Figure 1: Primary Si crystal in a T5 heat treated composite

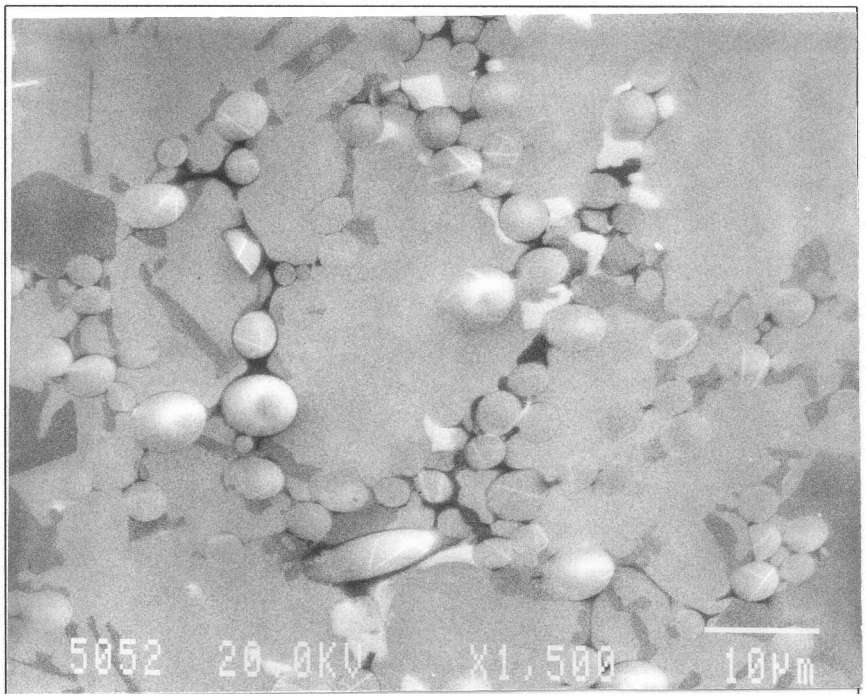

Figure 2: Mg_2Si precipitates surrounding the ceramic fibres.

Testing parameters

The fracture toughness was determined according to the standard procedure of ASTM E399. The specimens employed were CT type with a chevron notch, as shown in the A4.1 of the named standard, in order to avoid excesive crack front curvature reported in composite materials by other researchers [2]. Following the nomenclature of the above named standard, the selected dimensions of the specimens were W=20mm and B=10mm. The rest of dimensions were defined according to W and B. Figure 3 shows a scheme of the specimens and their dimensions. These specimens are in the limit of the small size allowance of the standard

procedure, which was due to the size of the composites, originally 20 mm thick. The testing speed employed was V=0.008mm/seg.

Totally 72 CT specimens (for different K_{IC} and fatigue studies) were machined and one of their surfaces polished to measure properly the notches produced in the precracking stage. The precracking consisted in fatigue 0,12/1,2 KN at 25 Hz. During this stage 33 specimens broke (45% of the specimens). The rest of the specimens required a number of cycles between 25.000 and 763.000. This is due both to the large heterogenity of the material and the small size of the specimens employed. All the precracks produced were valid for K_{IC} testing. Two trials were made to produce the initial acute nothes by rotative compresion-compresion fatigue, which presents smaller probability of catastrophic failure than the tensile fatigue. The results, however, obligued to reject this method because the cracks obtained did not grow uniformly across the thickness of the specimens.

Five specimens were tested at 20ºC, 75ºC, 150ºC, 200ºC and 250ºC. Both the precracking and the tests were performed in an Instron 8501 servohydraulic dynamic testing machine. The crack tip displacement was measured by an extensometer COD (Instron cat nº 2670-116) with 10 mm base and 4 mm maximum displacement. The high temperature tests were preformed in a heating cabinet Instron 3110. the specimens were prepared at room temperature, heating to the testing temperature at a rate of 75ºC/min, and maintaining at this temperature for 30 min prior to testing. For temperatures above 100ºC there was not available any commercial extensometer, and a special device was prepared to make the displacement measurement out of the heating cabinet. This device was produced in a light (density around 1,7 $g.cm^{-3}$), extremely high modulus and low CTE carbon-carbon composite from S.E.P. to guarantee a good transmission of the crack tip displacements, minimising the error due to the device deformations during the measurments. This device was tested in room temperature tests with well known materials (aluminium alloys) to check its reliability, prior to the tests performed in this study.

Fracture analysis

The fracture surfaces were all observed macroscopically, and microscopically in S.E.M. Crack path profile samples were metalographically prepared, and microspcopically observed. The initial crack lengths were measured and failure mechanism were discussed according to the observations.

RESULTS AND DISCUSSION

The precracks produced were valid according to the ASTM standard employed. The employment of Chevron like notches was crucial, since the works of Roebuck et al (178, 179) reported a great difficulty in the production of plane crack formts in particulate matrix composites, both in CT type specimens and in SENB specimens when the notches are simple. The results of the K_{IC} tests performed are gathered in table 3.

TABLE 3
TEST RESULTS

Temperature	K_{IC} (MPa.m ^{½})	Max. value	Min. value
20ºC	8.74	11.11	6.725
75ºC	9.87	11.49	9.04
150ºC	9.51	10.29	9.026
200ºC	9.20	10.13	8.023
250ºC	7.52	8.828	6.215

All the tests were valid according to the conditions of the ASTM standard employed except one of the experiments performed at 250ºC. The non valid experiment presented a Pmax/Pmin <1.1, and therefore the determined stress factor is K_Q and not K_{IC}.

A first glance to the results shows a brittle material with a very low K_{IC} value and a large scattering. Curiously, the mean value at room temperature is clearly smaller and the scattering is considerably higher than those at higher temperatures, until reaching 250ºC, where the mechanical behaviour of the material seems to fail. This behaviour is due to the effect of the matrix ductility in the range betwen 50ºC and 250ºC aprox. At room temperature the alloy employed in the composite is very brittle, and therefore the behaviour of the composite corresponds to a highly heterogeneuos brittle material. At higher temperatures, the alloy becomes more ductile. In this range the composite behaviour correspond to a typical composite of brittle rigid fibre in a ductile matrix, and therefore the fracture toughness increases and the scattering of the results decreases. As the temperatures reaches 250ºC, the alloy becomes more plastic, and the composite looses its brittle elastic macroscopic behaviour.

However, in spite of the low fracture toughness measured, the type of crack propagation mechanism is not that of a typically britlle material. The energy absorbed for catastrophic crack propagation is in this case of the same order than the energy absorbed for critical crack initiation, while in brittle materials almost all the fracture energy is spent in the critical crack initiation.

FRACTURE ANALYSIS

A first observation of the general fracture surfaces denotes a very high roughness, rarely observed in metallic materials (figure 3). The determination of the initial crack according to the features observed in these surfaces was not always easy, and an oil impregnation technique had to be employed.

The fracture surfaces and the crack profiles of the specimens tested in all the temperature range show a macroscopically brittle behaviour while microscopically the fracture mechanism combines brittle failure of the fibres and the intermetallic phases, and ductile deformation of the α-aluminium phase.

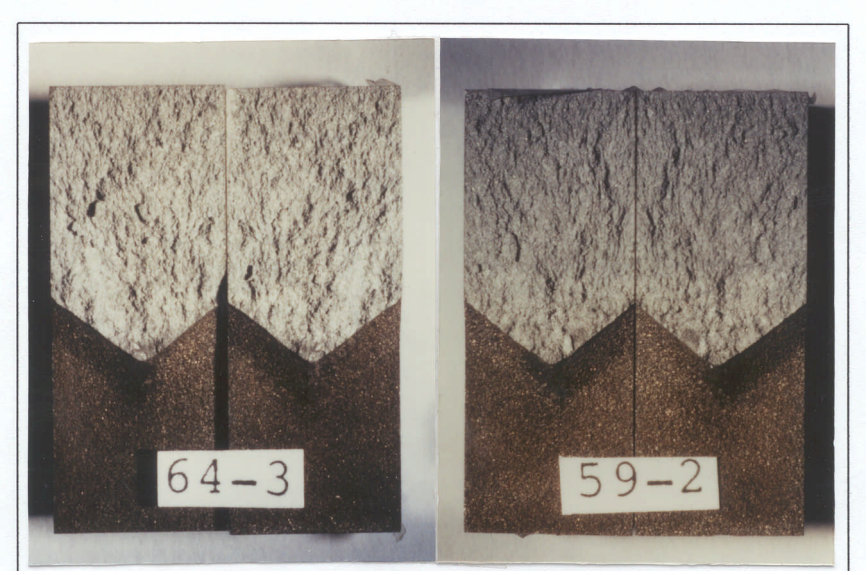

Figure 3: Fracture surface of a K_{IC} tested specimen at room temperature

Apparently, the degree of plastic deformation of the α-aluminium phase increases as the temperature increases. The quantification of this visual observation, however, in not possible due to the high complexity of the composite composition. The fracture surfaces show cracks growing by deformation and decohesion of the interface between the α-aluminium and the fibres and/or intermetallic phases. The profile of the cracks are formed by α-aluminium and fibres, while secondary phases are seen only scarcely. Both observations lead to a crack propagation mechanism of decohesion between the α-aluminium and the fibres mainly, and secondarily between the intermetallic phases and the fibres. Curiously, the adherence between the fibres and the alloy is very good in spite of being the crack progression mechanism, and the fracture does not show a delamination mechanism, but a decohesion due to deformation of the phases around the fibres, as can be observed in figure 4.

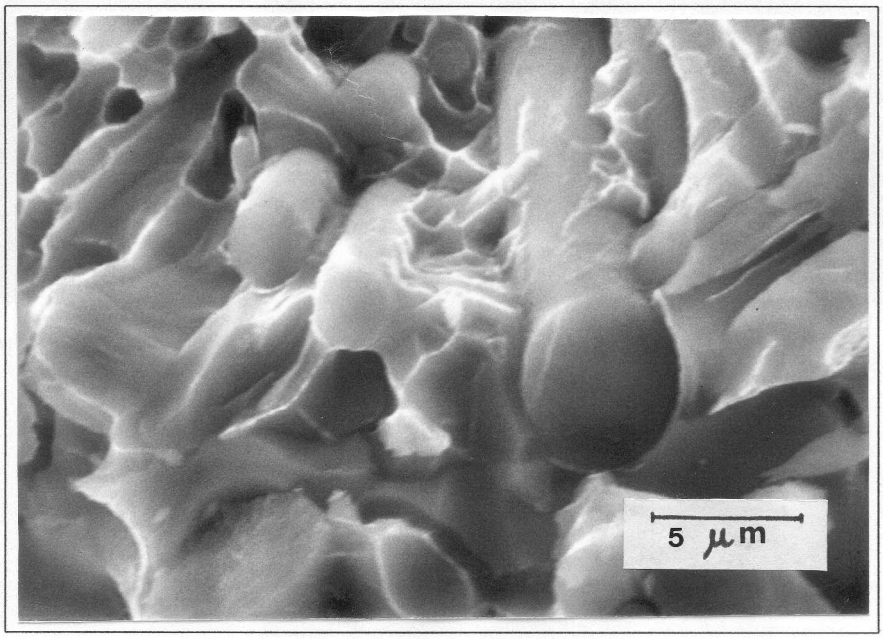

Figure 4: Fracture surface of a specimen tested at room temperature

As the temperature increases, the fibre-matrix bonding becomes weaker, and delamination and fibre pull-out appears sporadically, as can be seen in figure 5, showing the fracture surface of a specimen tested at 250ºC.

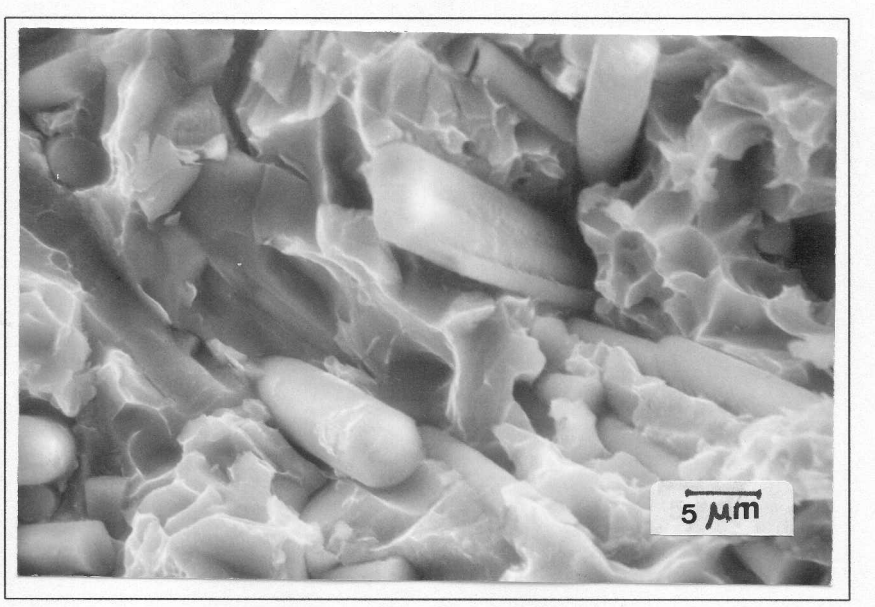

Figure 5: Fracture surface of specimen tested at 250ºC

CONCLUSIONS

The composite materials evaluated are macroscopically brittle and outperform better in a range of temperature between 75ªC and 200ºC than at room temperature.

The fracture toughness is very low, but the energy absorbed in fracture propagation is considerably larger than that shown by typically brittle materials

The fracture mechanism in the whole temperature range tested is decohesion between the α-aluminium and the fibres. Pull out and delamination are negligible until temperatures around 250ºC, when the composite fracture toughness decreases considerably.

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