

# FRACTOGRAPHIC STUDY OF TWO DAMAGE TOLERANT ALUMINIUM ALLOYS

A.M. Irisarri and G. Atxaga

Structural Integrity. Fundación INASMET.  
C. Portuetxe 12. 20009 San Sebastian

## ABSTRACT

The influence of second phase particles on the fracture behaviour of two high strength aluminium alloys (7475 and 7050) has been analysed. Fractographic study was carried out by scanning electron microscopy and energy dispersive spectrometry was used to identify the particles present on the fracture surfaces. Examination of the failed specimens revealed that the whole fracture surfaces were covered by ductile dimples pointing towards microvoid coalescence as operating mechanism and evidences the important role played by the second phase particles. Most of these particles were iron rich ones, varying the other elements, although mainly corresponded to the aluminium-copper-iron type. These observations showed the significant contribution of the iron content to the failure process and confirmed that the used approach to improve toughness by reducing the percentage of iron is a good practice and even more strict control of this element is recommended.

## INTRODUCTION

The call for improved economy in connection with high reliability is the overriding requirement for the development of aircraft. The structures should have a long life, low weight and minimum maintenance costs. To satisfy such properties, materials providing not only high static and fatigue strength but also greater resistance to crack propagation, high residual strength and good corrosion characteristics are required [1]. Aluminium alloys have satisfied these requirements best for more than 50 years and therefore make up for most of the aircraft structure. One vision of the subsonic commercial jetliner of about 2020 suggests an aircraft with a structure containing almost entirely of polymer matrix composites. However, the Boeing 777 was also predicted to be built of about 70% of polymer matrix composites, and yet is predominantly built in conventional aluminium alloys [2].

Only alloys from two systems, both precipitation hardenable, are serious candidate for aircraft structures; the aluminium-copper-magnesium (2XXX series) and the aluminium-zinc-copper-magnesium (7XXX family) systems. High strength Al-Zn-Cu-Mg based wrought 7XXX alloys are widely used for structural applications in aircraft industry, where fracture toughness and yield strength are of prime concern. Aircraft designers generally prefer 7XXX alloys for lower wing structures because of their high strength even though they have only moderate damage tolerance. The story of the development of aluminium alloys over the greater part of the last decades has been one of early, and generally quite successful, attempts to increase the static strength, followed by the realisation that this property is not necessarily the one of most importance. A high value of static strength is, of course, always desirable and the development procedures applied to any alloy cannot be allowed to sacrifice the mechanical strength to any considerable degree. However, almost all of the research

effort which has been directed towards aircraft alloys in recent years has been aimed at the enhancement of toughness, fatigue resistance and resistance to the various types of corrosion, even where this has involved some loss of strength [3]. This target was achieved by limiting the content of some accompanying elements, especially iron and silicon, and by optimisation of the entire manufacturing process from casting to the final product. Alloy 7475 is an example of this improvement in composition by more strict control of these impurities. This alloy develops strength levels comparable to those of traditional 7075 in similar tempers, but with increased toughness, being possible its use for lower wing skins and fuselages. Moreover, the necessity of higher through hardening properties lead to the development of new alloys in which chromium was replaced by zirconium as the element added to inhibit or retard the recrystallisation process. One of these alloys, available since the last seventies, with controlled contents of iron and silicon is the alloy 7050.

Three different types of second phase particles can be distinguished in heat treatable aluminium 7XXX alloys. The first one corresponds to the intermetallic particles formed during the solidification, due to the segregation of elements. These intermetallic particles can be hardly taken into solid solution and are virtually unaffected by heat treatment. A second class of particles, denominated dispersoids, is formed by solid state precipitation. They are constituted by chromium, manganese or zirconium, elements added to high strength aluminium alloys to modify the dynamic recovery and thus to control static recrystallisation in hot worked products, and other alloy elements such as aluminium, magnesium or copper [4]. The third type of particles consists in the very fine metastable precipitates that are responsible for most of the strengthening of the alloy. These precipitates are formed during the holding of the alloy, rapidly quenched from above the solvus temperature, at either room temperature (natural ageing) or elevated temperature (artificial ageing) A change in the number, size, shape and nature of these precipitates is observed during this process from near spherical particles denominated Guinier-Preston zones, through a transition  $\eta'$  semicoherent precipitate, and, finally, to the equilibrium incoherent  $\eta$  phase [5]. Logically these different particles must have a different influence on the mechanical properties of the alloy. Consequently, in order to develop improved materials, larger effort is needed in understanding how the microstructure affects the properties of these alloys.

The aim of this paper is to analyse the influence of second phase particles present in the microstructure of two high strength aluminium alloys on their fracture and fatigue behaviour.

## EXPERIMENTAL PROCEDURE

Materials chosen for this study corresponded to two 50-mm thick plates of two high strength aluminium-zinc-copper-magnesium alloys, 7475 and 7050, whose chemical compositions are given in Table 1.

TABLE 1  
CHEMICAL COMPOSITION OF THE TWO ALUMINIUM ALLOYS

Alloy	Zn	Cu	Mg	Fe	Si	Cr	Zr	Al
7475	5.6	1.6	2.4	0.09	0.10	0.20	-	Balance
7050	6.0	2.3	2.2	0.12	0.11	-	0.12	Balance

Alloy 7475 was in the T7351 condition, that is, solution treated after a high temperature soaking time to achieve a nearly homogeneous solid solution, rapidly cooled, stress relieved by stretching and two stage aged to obtain the required precipitation hardening. The strength achieved after this treatment is lower than the peak strength but possesses the advantages of higher toughness and better exfoliation and stress corrosion cracking resistance. The T7451 condition, used for the alloy 7050, consists in a very similar treatment but with stress corrosion resistance and mechanical strength intermediate between those of T7351 and T7651 conditions.

Tensile tests were carried out at room temperature on specimens machined from each plate at two different locations (1/4 thickness and middle of the plate) and with two different orientations (longitudinal and long

transverse). Fracture toughness tests were also performed at room temperature using 25-mm thick CT specimens machined at the same locations that the tensile test specimens, in the L-T and T-L orientations. A more detailed description of this experimental procedure was published in reference [6]. In order to facilitate the analysis of the influence of the various particles on the fracture behaviour of both alloys the average values obtained in all these tests are summarised in the Table 2.

**TABLE 2**  
MECHANICAL PROPERTIES OF THE TWO ALLOYS AT THE VARIOUS LOCATIONS AND ORIENTATIONS

Alloy	Position	Orient.	Y.S (MPa)	U.T.S. (MPa)	Elongation. (%)	K <sub>Q</sub> (MPa.m <sup>1/2</sup> )	K <sub>IC</sub> (MPa.m <sup>1/2</sup> )
7475	¼ thickness.	L-T	421	492	10.4	51	-
7475	¼ thickness	T-L	423	496	10.2	48	
7475	Centre	L-T	416	487	10.0	49	
7475	Centre	T-L	417	489	10.2	47	
7050	¼ thickness	L-T	441	510	11.2	37	37
7050	¼ thickness	T-L	441	510	10.1	34	34
7050	Centre	L-T	437	505	10.3	36	36
7050	Centre	T-L	437	507	10.3	34	34

No significant influence of either the location or the orientation of the specimens on the recorded values in the various tensile and fracture toughness tests was observed, being the values recorded on specimens machined at the centre of the plate just a little below than those recorded at ¼ thickness. Fracture toughness values in the L-T orientation are also very slightly higher than T-L ones. This result seem to contradict those reported by Bucci [7] who found significantly higher values in the tests carried out on L-T specimens. Most plausible explanation will be based in differences in the rolling process between the plates analysed in each study. Much larger differences were found between the results recorded for each alloy. Alloy 7050 possesses a higher strength than alloy 7475 but the last one is markedly tougher, to the point that 25-mm specimens are not thick enough to obtain valid K<sub>IC</sub> values according to the established requirements.

Fatigue tests were carried out on both smooth and notched specimens machined at the same locations that the tensile and fracture toughness ones. More detailed description of the testing procedure and the results obtained are given in another paper presented at this Conference [8].

After failure, the fracture surfaces of all these tensile, fracture toughness and fatigue specimens were examined in a scanning electron microscope to reveal the operating mechanism. Particles and phases present in these fracture surfaces were identified by means of energy dispersive spectrometry.

## RESULTS AND DISCUSSION

Metallographic examination of both alloys revealed that they were constituted by recrystallised grains, extremely elongated in the rolling direction because of dispersoid banding and unrecrystallised regions made up of very fine subgrains in which boundaries are decorated by hardening precipitate.

Scanning electron microscope examination of tensile and fracture toughness specimens revealed that the whole fracture surfaces were covered by ductile dimples, most of them with particles inside, pointing towards microvoid coalescence as the operating failure mechanism. No clear difference was observed between the specimens machined at the various depth locations or notch orientations. This observation is in very good agreement with the similarity in the results recorded in the mechanical tests As it is known these microvoids nucleate at regions of localised strain discontinuity, such as that associated with second phase particles. As the strain in the material increases, these microvoids grow, coalesce and induce the total failure. It is evident the important role played by second phase particles in this process as it is their fracture or decohesion to the

matrix which leads to preferential paths for crack advance and reduces fracture toughness [9]. Consequently, the identification of the nature of these particles could help to improve the fracture toughness of these alloys.

As it has been previously indicated this identification of the various particles was carried out by means of energy dispersive spectrometry. A second analysis of each particle was performed using low beam energy to minimise its penetration into the sample and the contribution to the spectrum from the matrix that surrounds the particles.

No difference in the fracture morphology was observed between specimens machined at the two depth locations or in the longitudinal and transverse orientations. This result is in good agreement with the very similar values recorded in both tensile and fracture toughness tests corresponding to any location or orientation.

A large number of these precipitates observed in the alloy 7475 correspond to iron rich ones, varying the other elements, even if most of them corresponds to the aluminium-copper-iron type. Even if significant changes in the amount of these elements from one particle to another were found their identification as  $\text{Al}_7\text{Cu}_2\text{Fe}$  seems reasonable. Figure 1 exhibit a micrograph of the fracture surface of one specimen, where the larger particles (marked as A in the figure) belong to this type and most of the remaining particles, which are present also possess a certain amount of iron in their composition, associated to different amounts of other elements such as aluminium, copper, chromium, magnesium, silicon or titanium.

These observations revealed the marked influence of the iron rich particles to the failure process. The role played by these particles in the failure process is even more significant when it is considered that the amount of iron present in the composition of the alloy 7475 is limited to very low levels (maximum 0.12%). The negative effect of the iron contents can be also seen in reference [7] by a comparison between the values of toughness of the 7075 and 7475, that possess a very similar composition but the maximum amounts of iron and silicon in the first one are limited to markedly higher levels (0.50%). It has been pointed out that these coarse particles do not contribute to strength, but because they are brittle, they either fracture or separate from the matrix when the local strain exceeds a critical value. Consequently, they provide preferential crack paths ahead of a crack. When load is applied the existence of these preferential paths decreases the energy needed to propagate the advancing crack [4].

These observations clearly indicate that decreasing the volume fraction of these coarse intermetallic particles increases fracture toughness and resistance to crack propagation. A decrease in the number of these coarse particles could be achieved by putting them into solution during the heat treatment of the alloy. However, these particles can be hardly solved during the solution heat treatment as temperatures must be limited to a safe level to avoid the consequences of overheating or low melting of the various eutectics identified in these alloys such as the ternary  $\text{Al-Mg}_5\text{Al}_8\text{-Mg}_3\text{Zn}_3\text{Al}_2$  whose melting points is only 450° C. It can be argued that the presence of copper in the composition of the 7XXX modifies the nature of the eutectics but the melting temperature of the quaternary  $\text{Al-CuMgAl}_2\text{-Mg}_3\text{Zn}_3\text{Al}_2\text{-MgZn}_2$  is just 25° C higher [10] and the risk of localised melting is still very high. Consequently, the used approach to improve the toughness of this alloy by reducing the iron content and the possibility to form these large particles, seems to be the only reasonable practice, and a more strict control of this element is encouraged.

Moreover, a small number of iron free finer precipitates were detected in the fracture surfaces. Even if low beam energy was used to identify these phases a certain signal coming from the surrounding matrix cannot be completely discarded in the analysis of this particles. Probably to detect the finest ones will require the use of transmission electron microscopy but it is considered that their contribution to fracture toughness is through the increase in yield strength [11] and no through a change in the topography of the fracture surfaces.

Fracture surfaces of the tensile and fracture toughness specimens of the alloy 7050 also showed the presence of a notable number of iron rich particles that have contributed to the failure. Even if the aluminium-copper-iron particles, similar to those in the alloy 7475, constituted a considerable percentage of the total, the nature of other particles is different. Most of the large particles found in this alloy possess aluminium, magnesium

and silicon. . The presence of coarse  $Mg_2Si$  particles in the microstructure of 7XXX alloys was in a paper by Sanders and Staley [4] but in the present case particles possess also a high percentage of aluminium in their composition. The presence of this element in the X-ray spectrum could proceed from the surrounding matrix but the absence of other elements, which are present in the composition of the matrix and do not appear in this spectrum, seems to indicate that, at least, the majority of the aluminium comes from the particles that possess a certain amount of this element in their composition. Figure 2 shows an example of the particles that are present in the fracture surfaces of the specimens of this alloy.

Moreover, these observations point towards a stronger contribution of silicon and magnesium in both the formation of the particles and the failure process of the alloy 7050 than in the previously analysed alloy 7475, although the contents of these elements in both alloys are near identical, as it is clearly seen in Table 1. No clear explanation for this different behaviour has been found but the lower toughness of the alloy 7050 could be attributed to the contribution of these particles to the failure process. There is no evidence that  $Mg_2Si$  or the iron rich particles enter the low melting eutectics [10] but the need to solve these coarse particles to achieve the required toughness could lead to localised melts. Once again the best practice is to prevent its presence in the material by a strict control of the contents of iron and, at least in this alloy, silicon.

Fractographic examination of the fatigue specimens revealed that fracture surfaces were mostly covered by striations than in those tested at highest stress amplitudes exhibited a wavy front and many secondary cracks (figure 3). Nevertheless, in some zones the presence of large particles was observed in both alloys. One example of these particles in the fracture surface of a fatigue specimen is shown in figure 4. These large particles have a negative effect on the fatigue performance, accelerating the failure process, as their decohesion to the surrounding matrix generates microvoids that facilitate the crack propagation. Most of the particles in the alloy 7050 are similar to those observed in the tensile and fracture toughness specimens. However, particles that accelerated the failure in alloy 7475 were formed by aluminium, copper, iron and chromium and those formed by aluminium, copper and iron that constituted the greatest number in the tensile and the fracture toughness specimens were scarcely detected.

Chromium is added to this alloy to form particles called dispersoids to modify dynamic recovery process and thus control static recrystallisation in hot worked products. To induce the desired effect this element must be retained in supersaturated solution and precipitate during subsequent heat treatment as  $Al_{12}Mg_2Cr$  particles [4]. However, neither the composition nor the size (clearly larger than that of the dispersoids) of the particles found in these fracture surfaces agree with this origin and most probably have been formed during the solidification process. No satisfactory explanation has been found for the difference between the particles that reduced the fracture toughness and those that accelerated the fatigue crack growth in the alloy 7475, as both type of specimens were machined from the same plate. In any case, the negative influence of iron in the performance of this alloy is confirmed as this element enters in the composition of these particles. In alloy 7050 zirconium is added instead of chromium to form the dispersoids and the nature of the particles must be different but, once again, a significant contribution of the iron, or silicon, rich particles to the failure process has been observed.

## CONCLUSIONS

- a. - No clear difference in the fracture morphology was observed between specimens machined at the two depth locations or in the longitudinal and transverse orientations. This result is in good agreement with the very similar values recorded in the mechanical tests.
- b. - Fractographic examination of tensile and fracture toughness tests revealed that the operating mechanism was microvoid coalescence in both alloys and the strong influence of the iron rich particles and encourages the recommendation of a strict control of the content of this element.
- c. - Fracture toughness in the alloy 7050 is lower than in the alloy 7475 for any location or orientation. This lower toughness could be associated with differences in the nature of the particles contributing to the fracture

process as not only aluminium-copper-iron large particles but also other formed by aluminium, silicon and magnesium contributed to the failure process.

d. - Fracture surfaces of the fatigue specimens were mostly covered by striations than in those tested at the highest stress amplitudes exhibited a wavy front and a large number of secondary cracks. Nevertheless, in some zones the presence of large particles, that accelerated the failure, was observed

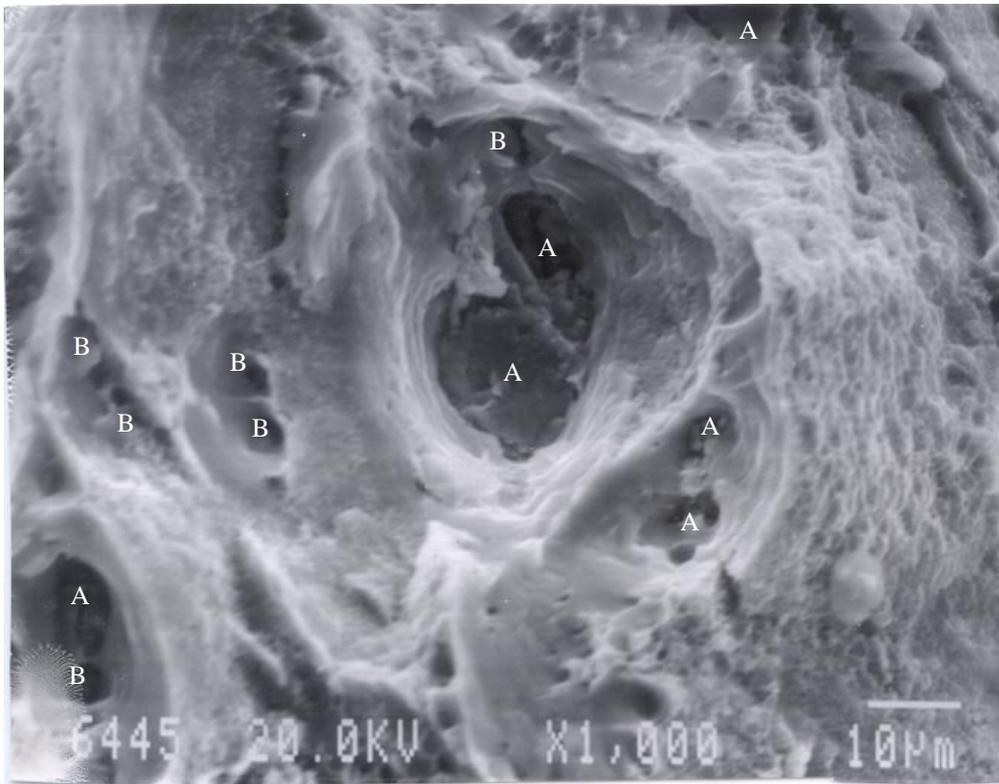
e. -Particles found in the fracture surfaces of the tensile fracture toughness and fatigue specimens of the alloy 7050 are near identical. However, the particles detected in the alloy 7475 fatigue specimens (formed by aluminium, copper, iron and chromium) differed from those observed in the tensile or the fracture toughness ones. No reasonable explanation has been found for this discrepancy.

## ACKNOWLEDGEMENT

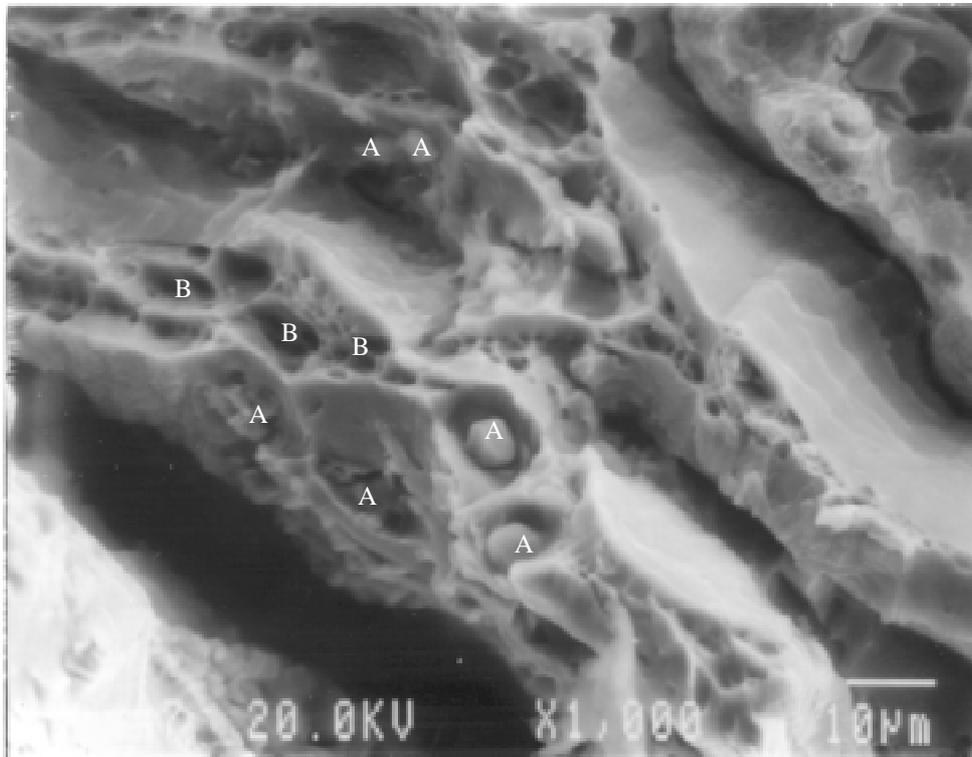
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## REFERENCES

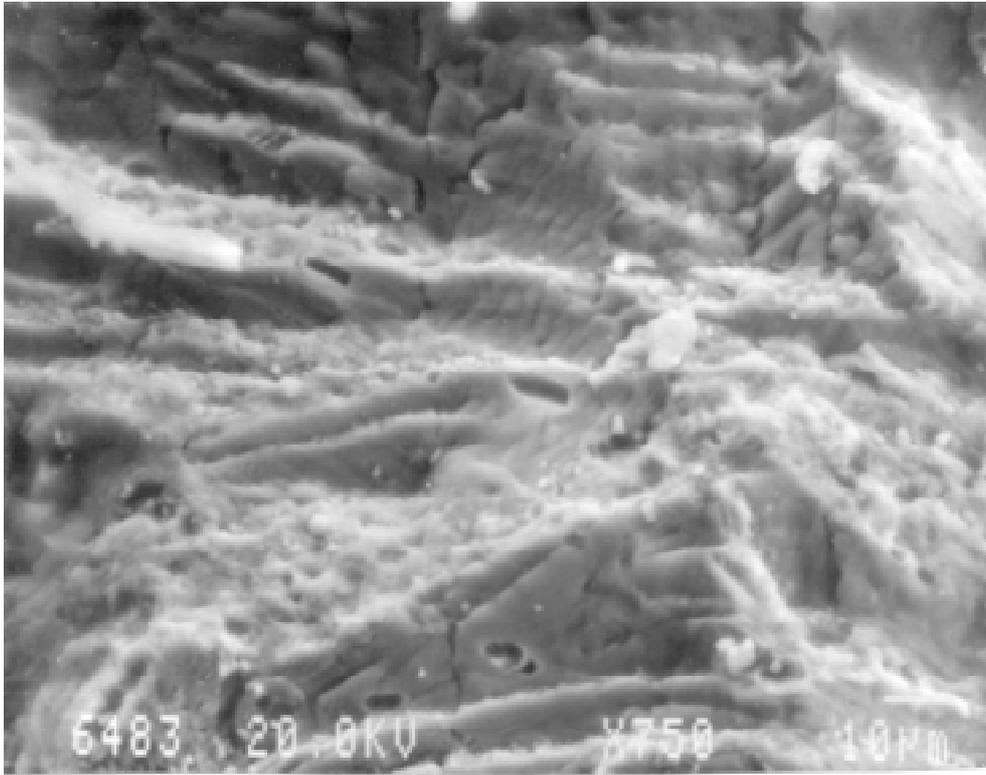
1. Rendings K.H. (1994) *Metallic Structures used in Aerospace during 25 Years and Prospects*. SAMPE '94 15eme Conference Exposition International Toulouse. Jun.
2. Staley J.T, Liu J. and Hunt W.H. (1997) *Advanced Materials & Processes* **152**, 4, pp 17-22
3. Crane F.A.A. and Charles J.A. (1985) *Selection and Use of Engineering Materials*. Butterworths pp 216-244
4. Sanders T.H. and Staley J.T. (1979) In: *Fatigue and Microstructure*. Meshii M. (Ed). ASM Metals Park Ohio, pp 467-522
5. Metallurgy of Heat Treatment and General Principles of Precipitation Hardening in *Aluminum, Properties and Physical Metallurgy*. (1984) ASM Metals Park. Ohio, pp 134-199
6. Atxaga G and Irisarri A.M. (1998) *Anales de Mecánica de la Fractura*. **15**, pp 192-197
7. Bucci R.J. (1979) *Engng. Fract. Mech.* **12**, pp 407-441
- 8 Atxaga G, Irisarri A.M, San Martín M and Plaza L.M. (2000) *Fatigue Behaviour of a Thick Plate of Aluminium Alloy 7475*. To be presented at this Conference.
- 9 Kerlins V. (1987) In: *Metals Handbook*. Ninth Edition. **12**. Fractography. ASM International. Materials Park Ohio, pp 12
- 10 Mondolfo L.F (1976). *Aluminum alloys. Structure and properties*. Butterworths, pp 842-882
- 11 Effects of Alloying Elements and Impurities on Properties. In: *Aluminum, Properties and Physical Metallurgy*. (1984) ASM Metals Park. Ohio. pp 200-241



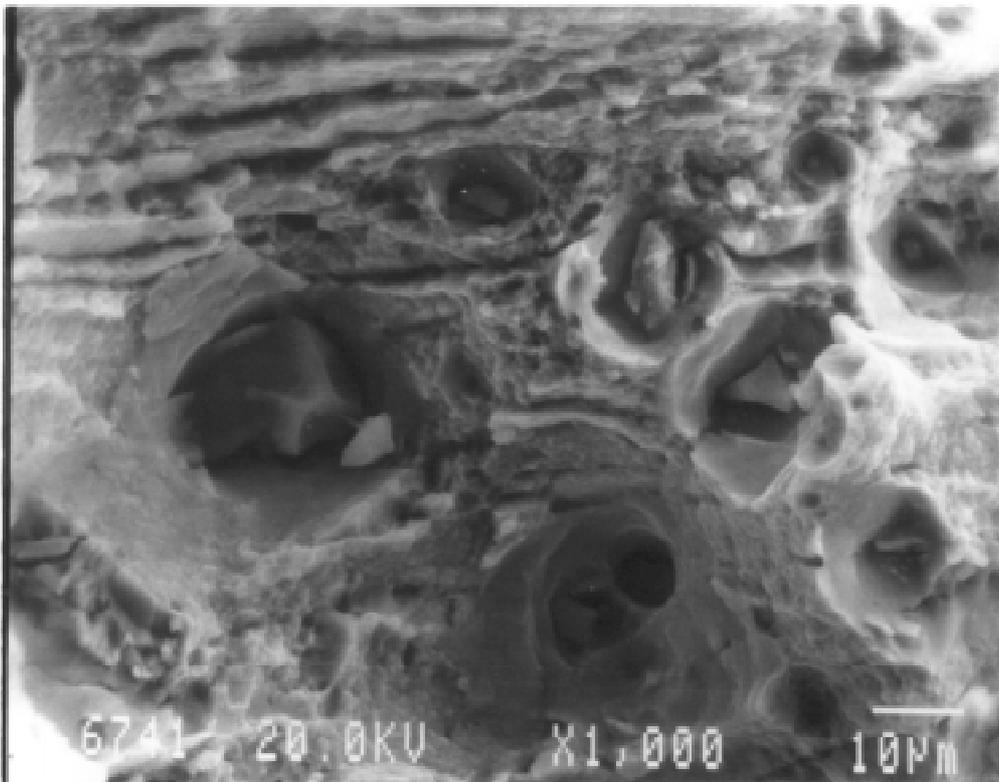
**Figure 1.** Micrograph Scanning Electron Microscope (X1000). Particles present in the fracture surface of one fracture toughness specimen of Alloy 7475. Particles marked as A correspond to aluminium-copper-iron, type, most of them with a composition near  $Al_7Cu_2Fe$ . Particles indicated as B are iron rich particles with various amount of aluminium, magnesium and silicon in their composition.



**Figure 2.** Micrograph Scanning Electron Microscope (X1000). Particles present in the fracture surface of one fracture toughness specimen of Alloy 7050. Particles marked as A correspond to aluminium-magnesium-silicon and B to aluminium-copper-zinc-iron type.



**Figure 3.** Micrograph Scanning Electron Microscope (X750). Fracture surface of one fatigue specimen Alloy 7050 covered by fatigue striations.



**Figure 4.** Micrograph Scanning Electron Microscope (X1000). Fracture surface of one fatigue specimen of Alloy 7475 showing fatigue striations and dimples with aluminium-iron-copper-chromium particles inside.