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TOUGHNESS AND HEAT TREATMENT RELATIONSHIP IN A 2091 ALUMINIUM ALLOY

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ABSTRACT: The 2091 alloy was tested to determine toughness levels with respect to heat treatment. A drastic decrease in fracture toughness was observed as a function of heat treatment. The occurrence of such a toughness drop was clearly related to fracture modes: Transgranular and intergranular precipitation and deformation modes were studied. The loss of grain boundary strength could be explained by the precipitation of the quasi crystalline phase T2

INTRODUCTION

Among the different aluminium lithium alloys developed the CP274 (2091) was chosen for this study. This material in the as-received state presents an adequate composition to provide high mechanical resistance and ductility. The composition of this alloy is shown in table 1. The as received alloy was in the form of rolled sheet 1.6 or 3 mm thick, solutionized at 525°C, water quenched followed by 2.5 % stretching and then aged at room temperature for 48 hours (T3). This material is fully recrystallized with a mean grain diameter of 35 μ m.

Element	Li	Cu	Mg	Zr	Fe	Si	Ti	Al
Weight(%)	1.9	1.85	1.2	0.07	0.07	0.02	0.02	94.72

Table.1: Alloy composition.

In this study the following heat treatments (H.T) were used :

symbols: T3 T8 A B C D E
H.T : as quenched 12h135°C 24h150°C 100h150°C 24h170°C 100h170°C 20h190°C

EXPERIMENTAL RESULTS**-Mechanical behaviour**

The complete results obtained from tensile tests and from toughness tests are gathered in table 2 for the main H.T.. It may be noted that an increase in yield stress and ultimate stress occurs as the H.T. approaches the peak aged and also that a drastic decrease in toughness is observed for the H.T. C. The toughness tests were conducted on CCT specimens ($w=200$ mm) following the ASTM B561 specification.

AGING TREATMENT	Y.S. (MPa)	U.T.S. (MPa)	ELONG. (%) ϵ_{max}	K_{IC} (MPa)	H _v
T3	298	407	18.7	82	120
T8	335	438	17	85	129.4
A	367	473	12	83	141.8
C	387	481	10	35	155.3

Table.2: Tensile and toughness results

-Fracture modes

The fracture associated with tensile tests remains essentially transgranular for H.T. T3 and T8, the fracture mode becoming intergranular for H.T. A and C. The fracture modes in toughness tests remain in the unstable propagation part. It may be noted however that H.T. A exhibits transgranular fracture mode in the stable propagation part. Plastic deformation is clearly revealed by straight striations on fractured grain boundaries.

-Transgranular precipitation

Aluminium lithium alloys containing a large amount of copper and magnesium are characterized [1] by the precipitation of $Al_{21}Li$ and Al_2CuMg phases. For each H.T. we have measured the average δ' diameter; all the results are plotted in fig.1

During the H.T.T3 and T8 we did not encounter S' phase. Nevertheless small GPB zones could not be detected with classical transmission electron microscopy (T.E.M.). GPB zones could induce small [100] direction streaks on the <001> diffraction patterns. In the as received condition (T3) this material exhibits a large number of helical or loop dislocations induced by quenching. A large quantity of linear defects could favour the heterogeneous precipitation of S'. Fig.2 shows the precipitation of S' phase on dislocations and loops, homogeneous precipitation is also encountered. A summary of our observations about S' phase for all the H.T. is shown in table 3 in which diffraction patterns are also included.

H.T.	T3	T8	A	B	C	D
S'	no	no	no	yes	yes	yes
HT.A						
HT.D						

Figure 1: Hardness versus δ' diameter
H.T. are included.

Table 3: S' precipitation various H.T.
diffraction patterns are included.

Figure 2: T.E.M. a) Bright field H.T. T3

b) Dark field and pattern H.T. E

-Intergranular precipitation

The structural study of grain boundaries for T3 and T8 H.T. shows that boundaries remain free of precipitation as illustrated in fig.3. Observations relative to H.T. A and C are shown in fig.4 where intergranular precipitates are clearly apparent. The crystallographic identification is shown on fig.5 where the typical five fold symmetry [2] of quasi crystalline phase is evidenced for H.T. C. It may be noted that there is no significant precipitate free zone for H.T. T3, T8, A and C.

Figure 3: T.E.M. Bright field on grain boundary H.T. T3

a

Figure 4: T.E.M. Dark field intergranular precipitation

a) H.T. A

b) H.T. C

b

Figure 5: T.E.M. Intergranular precipitate identification. a) Dark field and diffraction pattern H.T. A . b) Dark field and diffraction pattern H.T. C.

Figure 6.a: T.E.M. Bright field deformation modes ϵ_{max} H.T. T8

-Deformation modes

The study of deformation modes was carried out quantitatively and qualitatively. Quantitatively: results are summarized in table 4 in terms of volume fraction of deformation bands; these results were obtained from [3] by optical microscopy on polished tensile samples.

Qualitatively: by T.E.M., to show the slip character, for every observation we put a (111) slip plane close to the Bragg incidence, the diffracting vector was not parallel to the [111] direction, the observations are shown on fig.6.

		T3	T8	A	C
$\epsilon=5\%$	f_v	0.16	0.26	0.14	0.16
	σ	0.06	0.06	0.04	0.04
ϵ_{max}	f_v	0.31		0.32	0.30
	σ	0.07		0.07	0.05

Table.4: Mean volume fraction of deformation bands(optical microscopy), σ standard deviation.

Figure 6.b: T.E.M. Bright field , deformation modes, $\epsilon_{\max}$ H.T.C.

DISCUSSION

The trend which is emerging from the preceding results is a strong decrease in the toughness level at H.T. close to the peak aged condition. This toughness decrease is accompanied by a particular precipitation on grain boundaries and by a relative stability in deformation modes. The following discussion will focus on these microstructural aspects in order to explain the toughness evolution. To determine the role of this precipitate on the toughness level a study was conducted using fractography replicas of the fracture surfaces from tensile tests with heat treatment A and C. These replicas show, in fig.7.a and c, that the fractured grain boundaries bear the trace of the intergranular phases; the main parameters obtained from the replicas are shown on table 5.

Figure 7.a: T.E.M. Replica , fracture modes tensile test H.T.C.

Figure 7.b: T.E.M. Replica , fracture modes tensile test H.T.A

	SYMBOL	A	C
mean diameter of intergranular precipitates(A):	$d_p =$	570	800
mean distance between intergranular precipitates(A):	$D_p =$	1500	1500
mean diameter of intergranular dimples(A):	$C_d =$	1800	3400
mean distance between intergranular dimples(A):	$C_d =$	2500	4000
crack opening displacement(μm):	$\delta_c =$	160	25
plastic zone size(mm):	$R =$	15	25

Table 5

δ_c and R were calculated with the following relations:

$$\delta_c = 2K_{IC}^2 / \pi E \sigma_Y$$

$$R = 1 / \pi (K_{IC} / \sigma_Y)^2$$

Table 5 shows that whatever the toughness value the size of the plastic zone(R) and the crack opening displacement (δ_c) are substantially larger than the grain diameter. Furthermore, it appears that the parameters D_p and C_d are not correlated.

The examination of deformation modes has shown that the slip homogeneity increases with H.T. approaching the peak aged condition. Two main factors influence the slip character i.e. the size of δ' and the size of S' precipitates. For the complete H.T. the size of δ' remains smaller than the critical size favouring localized deformation bands. Nevertheless other factors could spread the slip: Firstly, the imposed strain level by multiplication of slip systems and secondly the presence of the more aged H.T. of GPB zones precursor of homogenous S' precipitates. It has been shown that GPB zones or S' precipitates in particular circumstances could be broken by dislocations (unpublished results: D.Khiredine, R.Rahouadj and M.Clavel). In our opinion the fact that homogeneous S' precipitates are scarce whatever H.T., could favour a by-passing process at high strains. Finally the relative homogeneity of slip is mainly governed by the amount of elongation avoiding the δ' and/or the GPB zone effect on localized deformation modes.

It is clearly shown that the distance between slip bands by T.E.M. and by optical microscopy does not correspond. The same results arise by observing slip band emerging at the surface of the grain boundaries. These observations indicate that among slip bands in the matrix only a few of them induce slip traces on the surfaces. We consider that the slip bands correspond to the first active slip planes during the strained process. If we want to understand the role of the slip character we must study the first stage of the deformation process i.e. for strains below 5 %. In other words optical quantitative metallography would appear to be an inappropriate method for the quantification of slip character if it is conducted at high strains. The difference observed in our results concerning the volume of slip

bands between 5 % strained and $\epsilon_{\max}$ is mainly induced by the number of deformed grains rather than the multiplication of slip traces in the strained grains. In the first stage of deformation, only the grains oriented for easy glide are plastically strained. Other grains, particularly those oriented for multiple slip, would be plastically strained at a higher level of strain. However, multiple slip induces dislocations crossing favoring a multiplication of slip bands. In fact, only few grains would be fractured by interaction between slip bands and boundaries, the others are fractured by dimple formation around intergranular precipitates. These results indicate that the deformation modes exhibited by the 2091 alloy for heat treatment T3, T8, A and C play in our opinion, a minor role in the toughness decrease.

From the previous results it clearly appears that toughness is mainly controlled by intergranular precipitation i.e. S (Al_2CuMg) for A H.T. and T2 for C H.T.. The grain boundaries for H.T. T3 and T8 remain free of precipitation. During the rupture process three stages are commonly considered [4], germination, growth and coalescence of dimples, using the preceding approach for intergranular failure; if we consider the size of grains and the extent of the highly strained zone ($2\delta_c$) at the crack tip, several grain boundaries are always highly strained in the plastic zone. In other words it is suggested that growth and coalescence processes do not actually intervene in the fracture process. In fact, initiation of voids is considered to be the main mechanism based on precipitate/matrix interface decohesion. The increase of ultimate stress noted from H.T. A to C could favour this decohesion. It would be noticed that coherency mismatch between S phase and matrix would be smaller than those encountered between T2 and the matrix, furthermore our observations demonstrate that the interface quasi crystal/matrix is mechanically weak. Careful examination of table 5 indicates that the mean distance between dimples does not correspond to the mean distance between precipitates, furthermore the growth rate of voids remains constant between A and C treatments and close to 4 considering the mean diameter of particles. This indicates that probably among the intergranular phases only the largest induced dimple formation [4][5]. The present discussion has been dealing only with intergranular failure as this phenomenon is obviously associated with toughness decrease. Meanwhile ductile fracture exhibited by H.T. T3 and T8 samples is a well known process [4][6] governed by intermetallic compounds.

CONCLUSION

The 2091 CP274 alloy exhibits a strong decrease of toughness (plane stress) associated with a transition of rupture modes from transgranular to intergranular. This toughness decrease appeared when the heat treatment approached peak aged conditions and is associated with the intergranular precipitation of the quasi crystalline phase T2. In fact deformation modes δ' and S' precipitates play a minor role in the decrease of grain boundary strength and it is suggested that T2 quasi crystalline precipitates on grain boundaries is mainly responsible for the drastic toughness decrease observed in this alloy.

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REFERENCES

- 1 P. SAINFORT, B. DUBOST and P. MEYER Invited Paper European Materials Research Society Fall Meeting Council of Europe, Strasbourg, France Nov, 26-28 1985
- 2 P. MEYER and B. DUBOST Aluminium Lithium III, Edited by C. BAKER, P. J. GREGSON S. J. HARRIS and C. J. PEEL p. 37
- 3 R. RAHOVADJ, J. MENIGAULT and M. CLAVEL, to be published in Mater. Sci. Eng.
- 4 G. G. GARETT and J. F. KNOTT, Metall. Trans., Vol 9A Sept 1978-1187
- 5 A. K. VAZUDEVAN, E. A. LUDWICZAK, S. F. BAUMANN, P. R. HOWELL, R. D. DOHERTY and M. M. KERSKER, Mat. Sci. and Techn. Dec 1986 Vol 2-1205
- 6 D. ALIAGA Thesis, Université de Poitiers, U.E.R., ENSMA, 29 Nov 1978