EFFECT OF MICROSTRUCTURE ON CRACK GROWTH IN A HIGH STRENGH ALUMINIUM ALLOY.

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Fatigue crack growth measurements in the low rate range and at near threshold level were made in air and in vacuum on a 7075 aluminium alloy in four aged conditions, viz. T351, T651, T7351 and 24 hours at 200 °C. Important differences were observed depending upon environment and aging conditions. The results obtained are discussed on the basis of microfractographic observations and specific mechanisms are proposed to describe each crack growth behaviour.

INTRODUCTION

Previous works have shown an important interaction between the effects of microstructure and environnement on fatigue crack growth (1, 2).

Experimental results presented in this paper were obtained on a 7075 aluminium alloy in four aged conditions : T351 (U.A.), T651 (P.A.), T7351 (O.A.) and a non standardized aging treatment of 24 hours at 200 °C (H.O.A.). New experiments carried out in vacuum are compared to results obtained previously in air (3).

Tests were carried out on CT specimens (W = 75 mm and thickness = 10 mm), using a servohydraulic machine equiped with an environmental chamber (11) providing a vacuum lower than 10^{-3} Pa.

The chemical composition and the tensile properties of the alloys used are presented in table I.

Crack growth which occurred in the long transverse/short transverse section, (grain size : 40 µm x 150 µm x 600 µm) was optically monitored with a travelling microscope with an accuracy of 0.01 mm.

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The propagation curves were determined by using a shedding method (serie of load drops) with a load ratio R = 0.1 and a frequency of 40 Hz.

RESULTS

Tests in air

The propagation curves da/dN vs ΔK in air (about 60 % R.H.) are illustrated in figure 1 for the four different aged conditions.

In the mid rate range the propagation appears to be independant of microstructure, but in the low rate range ($da/dN < 10^{-5}$ mm/cycle) a substantial effect is observed leading to different threshold levels. A change in crack growth behaviour is observed at a crack growth rate called $(da/dN)_k$ corresponding to a kink in the curves, except perhaps for the H.O.A. condition for which a more progressive evolution is observed without a well defined kink.

A detailed analysis of these curves was made in a previous paper (3) underlining the following points :

- for $\frac{da}{dN} > (\frac{da}{dN})_k$, in all cases the crack growth is essentially governed by the effect of water vapour adsorption, independantly of microstructure.

- for $\frac{da}{dN} < (\frac{da}{dN})_k$ there exists evidence of oxide deposits on the cracked surfaces which increase in thickness with aging and correspond to decreasing threshold level. Such a result is not consistent with the expected effect of the "oxide induced closure" mechanism proposed by Ritchie (4). Thus a new mechanism was proposed in ref. 3, by which the existence of an oxide layer could tend to inhibit the effect of water vapour at the crack tip as shown previously on a 2618 alloy (5), and leading to a crack growth process essentially controlled by the microstructure.

To get a better understanding of the effect of microstructure alone, a serie of tests was conducted in vacuum.

Tests in vacuum

In figure 2 the crack propagation measurements in vacuum and for each aging condition are presented in double logarithmic plots of da/dN versus ΔK. In the mid-range no variation between the overaged specimens and the P.A. one was observed, whereas the underaged alloy presents lower growth rates.

Below 10^{-5} mm/cycle the P.A. curve diverges from both the O.A. and H.O.A. ones and tends towards the U.A. curve.

The microfractographic observations made in the four cases showed considerable variations in fracture surface morphology with respect to the aging conditions :

- a very rough fracture surface consisting of crystallographic facets (Fig. 3), with some evidence of intergranular failure and ob branching, onserved an the U.A. specimen in all the explored rate range and on the P.A. specimen near threshold. Striations are observed on these facets as shown on a replica presented in the fig. 3 ; the interstriation is not coherent with the crack growth rate.

- a very flat and transgranular crack growth path, only marked by the wake of the intermetallic constituant particles of some µm in diameter (Fig. 4), observed on the overaged specimens in all the explored rate range and on the P.A. specimen only in the mid rate range (> 10^{-5} mm/cycle). Replica of the cracked surface near threshold shows evidence of a propagation mechanism controlled by the hardening precipitation (Fig. 6).

DISCUSSION

Tests in vacuum

The results obtained in vacuum show the existence of three crack growth behaviours depending upon aging conditions :

1 - 7075 U.A. contains G.P. zones consisting of small coherent precipitates which are easily shered by moving dislocations (6). The damage is characterised by the storing of dislocations in all the deformed volume in the absence of intrinsic trapping points (7). Crack growth resistance is expected to be improved due to a slow damage accumulation. In addition, a low Stacking Fault Energy (S.F.E.)restricts cross-slip (8). All the above facets are consistent with a predominantly planar slip mode and, in absence of any preferred path of fatigue crack advance, results in a chaotic fatigue fracture surface with large crystallographic facets as observed in fig. 3. The high resistance of the matrix containing G.P. zones might induce the occurrence of the observed intergranular fracture mode at favorably oriented grain boundaries and is consistent with a high K_{th} level.

2 - The overaged specimens contain larger precipitates (about 100 Å to 200 Å in size) which would resist shearing by moving dislocations (8), and they contain less Cu and Mg in solution which induces a higher S.F.E. favorising cross-slip (8). Therefore, a wavy slip mode can occur and a localized dislocation accumulation at the precipitate-matrix interfaces induces a preferential growth path controlled by the hardening precipitation.

All these are consistent with a very flat growth path and a faster crack growth than above. A less resistant matrix, without G.P. zones, is in accordance with a lower threshold level.

3 - In the P.A. aged condition the alloy contains G.P. zones plus small coherent plates (size 15 Å x 50 Å) (10) leading to the very high strength characteristics of this alloy.

A mixed behaviour can occur in this case :

- at low rate crack advance occurs step by step and the strain is localized to one shear direction (9). A crystallographic mode of failure, as described in the U.A. condition,can occur with a comparable threshold level.

- At higher rates ($> 10^{-5}$ mm/cycle), the crack tip plasticity affects many grains and crack advance at each cycle is higher than the precipitate spacing (about 100 Å). In addition this precipitation could favorise the occurrence of cross-slip (8). A wavy slip mechanism could occurs as in the overaged alloys.

This analysis is consistent with the crack growth data and the microfractographic observations presented above.

Tests in air

Crack growth behaviour near threshold in air can be discussed in the light of the above analysis of the tests performed in vacuum.

Complementary observations were made after replicating the fracture surfaces. This technic gives interesting informations since the oxide layer was stripped out, and thus it was possible to examine the oxide-metal interface morphology instead of the oxide layer alone as observed in the scanning microscope.

These observations indicate the existence of comparable features as in vacuum for each aged condition. Figure 7 shows striations on the facets of the crystallographic fracture surfaces of the U.A. or P.A. specimens which are assumed to correspond to step by step advance as in vacuum.

On the overaged specimen a flat growth path is observed corresponding to a mode of failure controlled by the precipitation hardening as in vacuum. However, the replica presented in fig. 8 shows a pronounced decohesion process at the interface of the M precipitates (10) which suggests a residual effect of environment in this case, which is not observed on the O.A. specimen and is in accordance with the crack growth data (Fig. 1 and Fig. 2).

This analysis of the crack behaviour near threshold in air is consistent with the mechanism previously suggested (3,5) consisting in a more or less pronounced inhibition of the environmental effect of water vapour adsorption in this rate range.

CONCLUSIONS

On the basis of the previous studies reviewed and of the results reported here and in (8), the following conclusions can be drawn about the crack propagation in the 7075 aluminium alloy in few different aged conditions.

In vacuum, three different behaviours can be distinguished :

- a planar slipping transgranular mechanism occurring in the G.P. zones observed in the U.A. alloy.

This mechanism results essentially in chaotic fracture surfaces with large crystallographic facets and some intergranular failure.

- a wavy slipping transgranular mechanism occurring in the presence of large hardening precipitates as observed in the overaged alloys and leading to faster crack growth, and, in absence of G.P. zones, to a lower threshold level.

- in the P.A. alloy both the above mentioned mechanisms occur: the first one in the G.P. zones area at low crack growth when the strain is localized to one shear direction and when the crack advance occurs step by step ; the second one in the mid-rate range when the mean crack advance at each cycle becomes higher than the M' precipitate spacing and when the strain at the crack tip affects many grains.

In air, for $da/dN > (da/dN)_k$, i.e. before the occurrence of oxide thickening observed for tests conducted at low load ratio, crack growth is essentially governed by the effect of water vapour adsorption independantly of microstructure (3).

For $da/dN < (da/dN)_k$ this effect tends to be inhibited by crack closure and oxide thickening so as mechanisms comparable to the ones observed in vacuum can govern crack growth near threshold in each aging condition.

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TABLE 1 - Composition (1-a) - Mechanical properties (1-b)

1-a - Composition

Zn	Mg	Cu	Cr	Fe	Si	Mn	Ti
6.00	2.44	1.52	0.20	0.16	0.07	0.04	0.04

1-b- Mechanical properties

Ref	Thermal treatment		$R_{0.2}$ (MPa)	R_m (MPa)	A%
UA	T 351	aging at room T°C	458	583	10.6
PA	T 651	24 h à 120 °C	527	590	11.0
OA	T 7351	6 h à 107 °C + 24 h à 158 °C	470	539	11.7
HOA	higly overaged	24 h à 200 °C	234	338	14.4

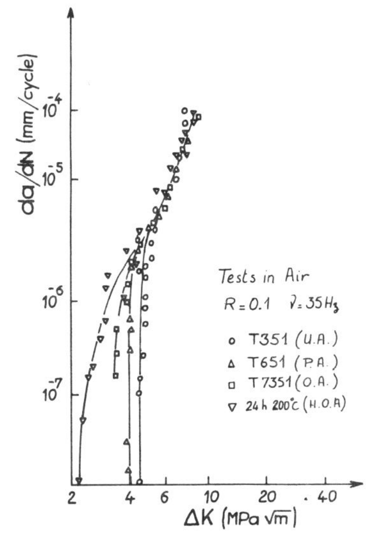

Figure 1 da/dN vs ΔK curves in air

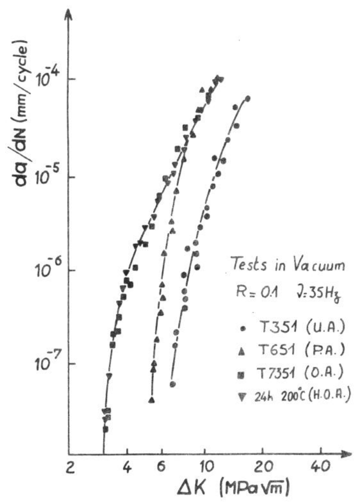

Figure 2 da/dN vs ΔK curves in vacuum.

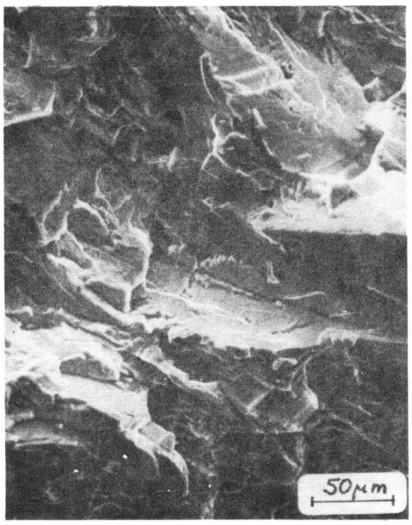

Figure 3 UA specimen in vacuum da/dN ≃3 10^{-6} mm/cycle

Figure 4 HOA specimen in vacuum da/dN ≃ 2 10^{-6} mm/cycle.

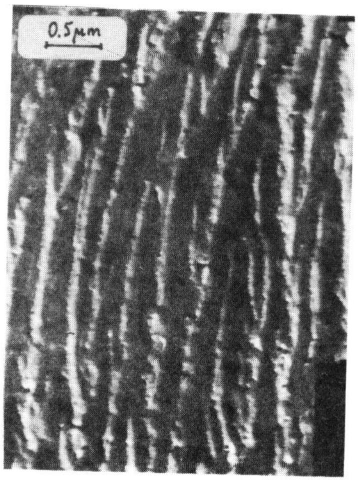

Figure 5 U.A Specimen in vacuum replica near threshold

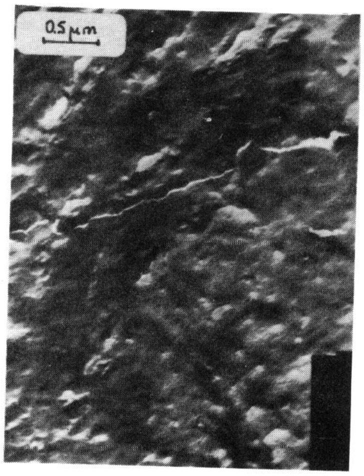

Figure 6 HOA specimen in vacuum replica near threshold

Figure 7 UA specimen in air : replica near threshold.

Figure 8 HOA specimen in air : replica near threshold.