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CONDITIONS LEADING TO LOCALIZED PLASTIC DEFORMATION AND FRACTURE IN SLIP BANDS DURING FATIGUE LOADING OF AGE HARDENING ALUMINUM ALLOYS

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INTRODUCTION

Aluminum alloys of the age-hardening type which can be heat treated to high unidirectional strength values have been developed over the last several years. Under cyclic loading conditions, however, because of the repeated shearing of the hardening precipitates in these alloys, localized micro-structural instabilities develop, leading to softening and concentration of deformation, in a few discrete slip bands [1, 2, 3]. In polycrystalline materials, crack nucleation and initial propagation occurs in such bands in grains lying at the surface; 50 - 80% of the total fatigue life is occupied by this so called stage I cracking [4]. In order to be able to understand and control crack propagation during stage I, it is important to know the factors which lead to localization of deformation in a few well-developed slip bands. This paper reports the results bringing out the influence of the type of alloying element(s), temperature and environment on the tendency to deformation and fracture in slip bands.

EXPERIMENTAL METHODS

In order to study the influence of the type of alloying element(s) and temperature on deformation and fracture behaviour, two types of fatigue experiments have been carried out with single crystals of two high purity alloys, Al-5wt%Zn-1wt%Mg and Al-4wt% Cu. The fatigue loading was always of the symmetrical tension-compression type.

1. Unnotched cylindrical specimens have been fatigued at a frequency of 0.33 Hz under constant plastic strain amplitude conditions ($\Delta\varepsilon_{pl} = 0.1\%$); the aim was to investigate the nature of the damage introduced by cyclic loading and to understand it in terms of the observed cyclic work hardening and softening behaviour. The AlZnMg and AlCu alloys were aged to produce large GP zones (1.5 nm radius) and θ'' precipitates (40 nm diameter and 2 - 3 nm thickness) respectively. The experiments were carried out at 295 and 77 K.

2. Flat single-edge-notched single crystals have been fatigued at 50 Hz under constant stress amplitude conditions. The principal aim was to investigate the crack propagation modes and velocities. The experiments were carried out at 295 and 77 K, the test environment at 295 K being dry nitrogen gas with a dew point of about 90 K.

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The influence of test environment on deformation and fracture behaviour was studied using single crystals of the AlZnMg alloy aged to maximum hardness. In this condition small η' particles are present; crack propagation experiments were carried out at room temperature with single-edge-notched crystals. Two test frequencies (50 and 5 Hz) were used, the test environment being either dry nitrogen gas or laboratory air.

EXPERIMENTAL RESULTS

Influence of the Type of Alloying Element(s) and Temperature

Experiments of Type 1

Figure 1 shows for the AlZnMg and AlCu alloys at 295 and 77 K the average stress reached in each cycle as a function of the number of cycles accumulated, which in the present case is directly proportional to the cumulative plastic strain. The average stress is defined as the average of the stress amplitudes in tension and compression in a given cycle. All the curves show an initial hardening stage. The experiment with AlCu at 77 K was interrupted while the hardening was still going on. In the other three cases, where the experiment has been continued well beyond the hardening stage, there ensues a softening stage. No well-defined or extended saturation stage was found to precede the softening stage. Thus, behaviour of AlZnMg and AlCu in the heat treatment conditions investigated here is different from that of a pure face centred cubic metal like copper, where over a wide range of plastic strain amplitudes an extended saturation stage is observed, enabling one to derive a *real* cyclic stress-strain curve.

As can be seen from Figure 1, in the fatigue tests carried out at room temperature fatigue softening was observed in both AlZnMg and AlCu. However, in the AlCu alloy the onset of softening is preceded by a higher amount of work hardening (75 MN/m^2 as compared to 57.5 MN/m^2 in AlZnMg) and a larger value of cumulative plastic strain (825 cycles to reach maximum in average stress as opposed to 185 cycles in AlZnMg). At 77 K the AlZnMg alloy showed a softening stage after about 900 fatigue cycles, whereas in AlCu no softening could be detected even after a very large number (4000) of cycles. In both alloys the amount of work hardening at 77 K is considerably higher than at room temperature (295 K). In addition the number of cycles required for the average stress to reach a maximum is higher at 77 K. In AlZnMg, for example, the maximum is reached after 185 and 900 cycles during fatigue at 295 and 77 K respectively.

After the fatigue loading history shown in Figure 1, metallographic observations were carried out on the original electropolished surfaces and on (101) surfaces produced by spark cutting perpendicular to the primary slip direction in order to characterize surface and bulk deformation respectively. In the latter case the AlCu samples were artificially aged to cause preferential precipitation at the deformation bands, and then the (101) surfaces were electropolished and etched to make the deformation visible. No prior artificial aging was necessary to make deformation in AlZnMg visible. Both surface and bulk deformation studies have shown that deformation at 295 K is concentrated in AlZnMg in a few discrete slip bands, whereas in AlCu the slip band density is considerably higher. Figure 2a and 2b show the results obtained with the (101) surfaces. Slip bands form in AlZnMg at 77 K also, but the bands forming at 295 K were found to be fewer, more discrete and well-developed than those forming at 77 K,

(Figure 2b and 2c). The same differences have been observed on the original electropolished surfaces. In AlCu at 77 K no evidence for any localized deformation could be obtained, even though fatigue was continued to much higher cumulative plastic strains (corresponding to 4000 cycles). Observations with both original electropolished surfaces of the cylindrical crystals as also (101) sections indicated that deformation remains relatively homogeneous and dislocation-rich bands do not form even after 4000 cycles.

Experiments of Type 2

Figure 3 shows schematically the crack propagation modes observed in the two alloys. Indicated are the values of stress amplitude σ_a (as a fraction of the yield stress of the material at the temperature at which experiment is being carried out σ_0) that are required to produce a steadily propagating crack, about 0.2 mm long, in about half an hour. At room temperature, the resistance to slip band crack nucleation and propagation in AlZnMg is very low; the crack path lies, from the nucleation stage until the final unstable fracture occurs, along slip bands. In AlCu at 295 K, crack nucleation and propagation occurred in slip bands, but the propagation velocities were markedly lower than those measured in AlZnMg alloy under comparable loading conditions. Furthermore, in AlCu an early macroscopic deviation of the crack path from the primary slip band occurs.

At 77 K slip band crack nucleation was found to be difficult, and in both alloy systems crack propagation occurred in stage II (i.e., normal to the loading axis). In order to study the influence of temperature on crack *propagation* in slip bands, a slip band crack was produced at 295 K in AlZnMg, and the test was continued at 77 K at the same value of stress amplitude. Propagation at 77 K continued to occur in what can be still characterized as stage I, but the propagation velocities were roughly four times smaller than those at 295 K. The fracture surface produced at 77 K, even though broadly lying in the crack plane produced at 295 K, is considerably rougher. It exhibits a larger number of steps and at places appears quite rugged.

Influence of Test Environment

The test environment in combination with the test frequency was found to exert a strong influence on the tendency to slip band cracking, and Figure 4 illustrates this effect for the case of AlZnMg. When the frequency is changed from 50 to 5 Hz during the crack propagation test, the propagation mode remains uninfluenced, if the test environment is dry nitrogen. Further, crack propagation velocity per cycle and the fracture surface appearance at the two frequencies are identical. (The appearance of fracture surface produced at 50 Hz in dry nitrogen environment has been discussed in detail in two earlier publications [5, 6]). However, in laboratory air environment, soon after changing over to 5 Hz, stage II cracks nucleate which then grow preferentially at the expense of stage I crack. Thus, the stage I fracture surface produced in laboratory air is limited in extent and in addition has a distinctly different appearance than the immediately preceding stage I surface produced in dry nitrogen. On the stage II surfaces produced in laboratory air one sees fatigue striations and also regions showing coalesced dimples.

DISCUSSION

The present experiments clearly show that the type of alloy additions and the test temperature play an important role in determining the deformation and fracture behaviour of age-hardening aluminum alloys. In AlZnMg aged to produce large GP zones, deformation at room temperature is localized in a few, well-developed slip bands. Changing the alloying elements from Zn+Mg to Cu, and lowering the temperature have qualitatively the same influence of homogenizing the deformation. The crack propagation experiments show that such a homogenization of deformation is also associated with a decreased tendency to slip band cracking and an increased tendency for stage I crack to change over into stage II. Thus, it appears that the explanation for the observed influence of the type of alloying element(s) and temperature on crack propagation modes can be sought in the intrinsic deformation behaviour of the materials; it also follows that deformation strongly localized in a few well-developed slip bands is a pre-requisite for slip band (stage I) cracking to occur on a macroscopic scale.

At lower frequencies slip band cracking is suppressed in laboratory air environment, and non-crystallographic stage II cracks grow preferentially. Similar result is obtained by Hockenhull and Monks [7] with an age-hardened Al-2.5Cu-1.5Mg alloy of commercial purity. Based on the analysis made in the preceding paragraph, it appears that the nature of the plastic zone at low frequencies can be different in laboratory air and dry nitrogen environments. Essentially a planar slip band lying at the crack tip in the crack plane is expected to constitute the plastic zone in dry nitrogen; development of such a slip band seems to be hindered due to the reaction of corrosive species present in the laboratory air environment with the fresh Al surface produced at the crack tip. This effect is not seen at high frequencies, indicating that time dependent steps are involved in the reaction between the material and the environment. One can think of several time dependent processes such as

i) arrival of the reactive species in the environment to the crack tip,
ii) adsorption of these species on the freshly generated aluminum surface at the crack tip, and
iii) diffusion of hydrogen into the plastic zone at the crack tip.

While the exact mechanism by which the material environment interaction influences the crack propagation mode is not known from the present experiments, it is clear that the time available in each cycle for the time dependent processes to occur is smaller at higher frequencies. Correspondingly, the probability that the crack propagation mode remains uninfluenced is higher at higher frequencies and this is in agreement with the present observations. The results also show that frequency alone has no influence on deformation and fracture in slip bands at room temperature in the frequency range 50 - 5 Hz. Wei [8], on the basis of careful analysis of experimental work of several different authors on stage II fatigue crack propagation in aluminum alloys has also reached the same conclusion viz. there is no intrinsic frequency effect during fatigue of Al-alloys over this frequency range.

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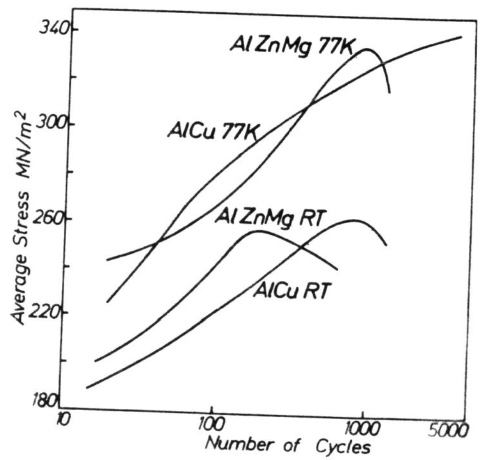
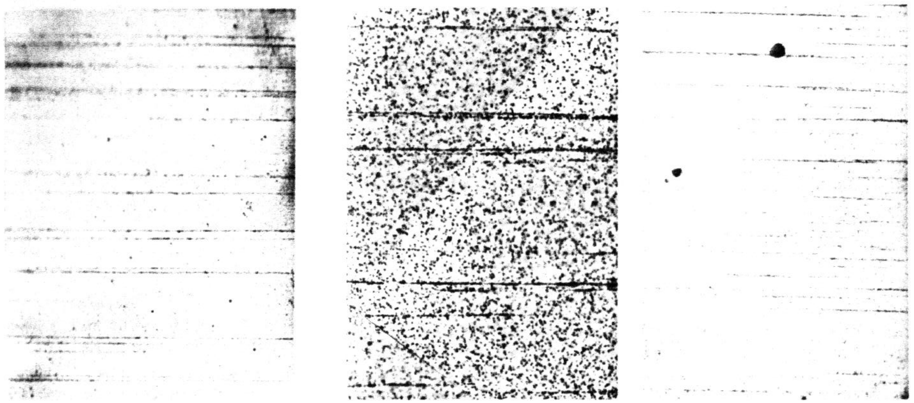


Figure 1 Cyclic Stress-Strain Curves



70X (a)

70X (b)

70X (c)

Figure 2 Slip Band Distribution as Observed on (101) - Sections Through Crystals of

(a) Al-Cu, Fatigued at Room Temperature
(b) Al-Zn-Mg, Fatigued at Room Temperature
(c) Al-Zn-Mg, Fatigued at 77 K

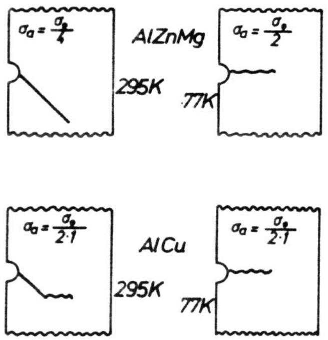


Figure 3 Schematic of Crack Propagation in Al-Zn-Mg and Al-Cu at 295 and 77 K

σ_a = Load Amplitude to Produce about 0.2 mm long crack in about 30 minutes

σ_o = CRSS

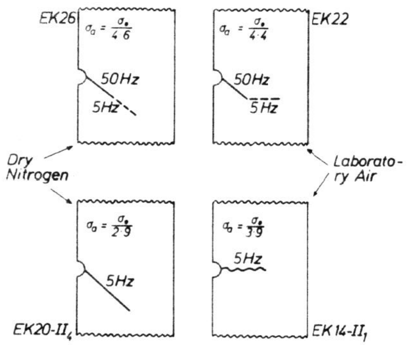


Figure 4 Frequency Environment Interaction During Fatigue Crack Propagation in Al-Zn-Mg