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Acoustic Emission from Al-3%Mg Alloy Deformed at Room Temperature

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Acoustic emission (AE) measurements during tensile tests of Al3%Mg polycrystalline alloy at room temperature have been performed. The relation between AE parameters and deformation characteristics has been studied. The material exhibited the B-type of the Portevin-LeChatelier (PLC) effect (serrated yielding). The results obtained are discussed in terms of existing models of the PLC effect.

1. Introduction

Many alloys (for example Al-Mg, Cu-Zn, Ti-O, steels) exhibit the dynamic strain ageing phenomena if deformed at suitable temperatures and strain rates. The most spectacular effect of these phenomena is related to an inhomogeneous fashion of the plastic deformation called Portevin-Le Chatelier (PLC) effect which is characterized by the appearance of serrations on the stress-strain curve.

The nature of the PLC effect has not yet been fully explained. Existing models can be divided into two groups.

In the first group of models the role of the diffusion of solutes is emphasized. Early models using mostly the Cottrell model [1] are treated for example by Hirth and Lothe [2], McCormick [3] and van den Brink et al. [4]. But the role of the bulk diffusion seems to be overestimated in these models (see the papers of Yoshinaga and Morozumi [5, 6] where a detailed calculation of the drag stress exerted on dislocations by bulk solute atmosphere is given). Alternative explanations were also published. According to Mulford and Kocks [7] mobile dislocations drain solutes by pipe diffusion from forest dislocations while they are waiting for thermal activation at the forest dislocations. Hong [8] proposed that the deformation stress contains a contribution related to dynamic strain ageing. Recently, Balík and Lukáč

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[9–11] have published a new model emphasizing the role of the pipe diffusion in dislocation cores. On the basic idea that solutes can be drawn by gliding dislocations authors [9–11] have shown dislocation cores may exist in two states called the enriched core and the saturated core. The corresponding stress-dislocation velocity dependences are shown in Fig. 1. The state of enriched core is typical for

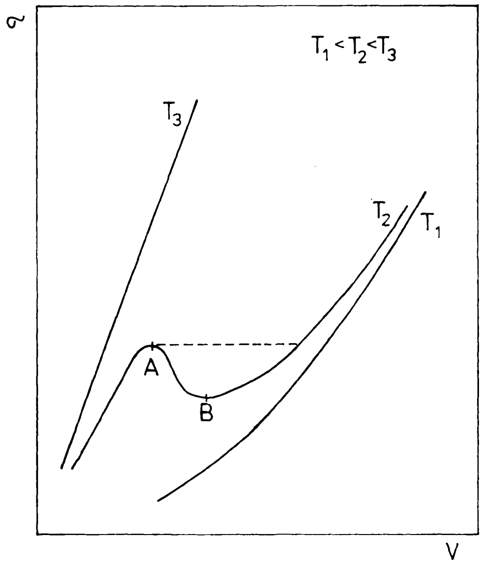

Fig. 1. Schematic stress-dislocation velocity dependences in materials exhibiting dynamic strain ageing. T_i, $i = 1, 2, 3$ are the absolute temperatures.

low temperatures when thermally activated motion of solutes can be neglected (T_3). The state of saturated core appears at high temperatures (T_1). At temperatures of $0.2-0.5T_m$, T_m is the melting point, the region of the common existence of both states may appear and the core of gliding dislocation can alternate between them passing a region of the negative strain rate sensitivity (T_2).

In the second group of models the diffusion is not presumed to play main role. Korbel et al. [12] explained the PLC effect to arise from the intrinsic nature of collective motion of dislocations. Onodera et al. [13] analyzed the influence of short range ordering realized by small precipitates created during the deformation. However, Pink [14] has shown that precipitates rather prevent the PLC effect than initiate it.

Tabata et al. [15] studied the PLC effect in situ by HVTEM and concluded the onset of the PLC effect to correspond to sudden collective motion of screw dislocations piled up on dislocation tangles.

The acoustic emission (AE) method has been often used to investigate the PLC effect. Results puslished up to 1987 are discussed in the review of Heiple and Carpenter [16]. More recent results were published by Cáceres and Rodriguez [17] and Polakovič et al. [18]. These results were discussed rather on the basis of the second group of the PLC effect models.

The aim of this paper is an attempt to discuss our AE results in terms of existing models of the PLC effect.

2. Experimental Procedure

A sheet of Al-3%Mg was used for this study. The preparation of the sheet is described in [19]. The average grain size was 20 μm. Tensile samples with the gauge dimensions 75 mm × 15 mm × 1 mm were cut parallel to direction of rolling.

The samples were deformed on the TIRA TEST 2300 testing machine with constant crosshead speed at room temperature. Several strain rate were selected ($\dot{\varepsilon} \in (2.2 \times 10^{-6}, 1.4 \times 10^{-3})\ s^{-1}$). The deformation was measured by a commerical extensometer the gauge base of which was shortened to 3.0 mm using special pieces.

The Dunegan-Endevco AE facility consisting of S9204 AE transducer (resonant frequency 150 kHz) and a D/E 3000 AE analyzer was used to measure cumulative parameters and amplitude distributions. Total gain was 75 dB and threshold level 35 dB.

3. Experimental Results

The true stress-true strain curves are shown in Fig. 2. The onset of the PLC effect is seen to begin later with increasing strain rate. A substantial deformation hardening

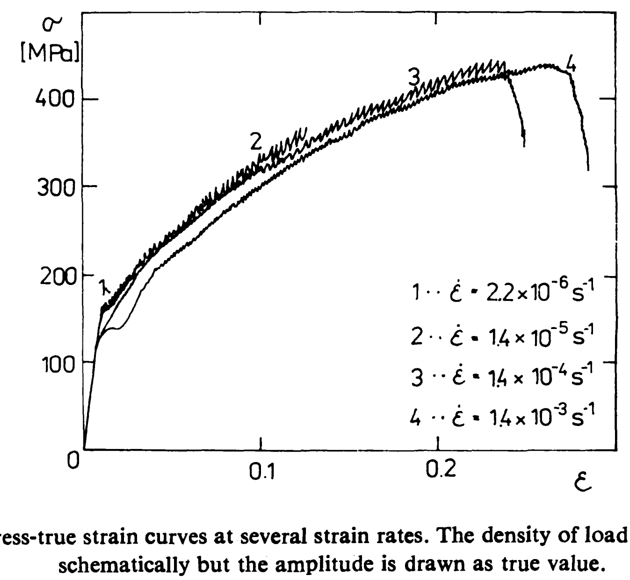

Fig. 2. True stress-true strain curves at several strain rates. The density of load drops is indicated schematically but the amplitude is drawn as true value.

was observed in all curves. Note the negative strain rate sensitivity in all samples (this behaviour was reported also by other authors [7, 9, 10]).

As it can be seen from the strain dependence of the count rate (Fig. 3), the count rate was observed to increase with increasing strain rate in the region preceding serrated yielding on contrary to the serrated yielding region where an increase of strain rate led to a decrease of the count rate. Note the oscillatory character of the dependence at the lowest strain rate. This corresponds to variable strain density

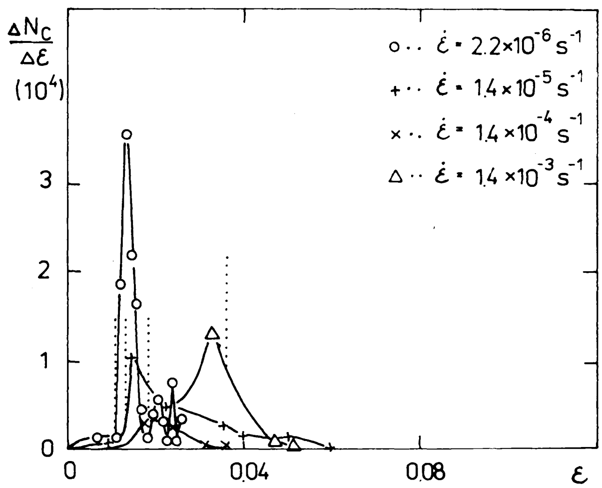

Fig. 3. Count rate-strain dependencies. The count rate was derived from total number of counts N_C as $\Delta N_C/\Delta\varepsilon$, where $\Delta\varepsilon = 0.001$. The onset of the PLC effect is indicated by dotted line.

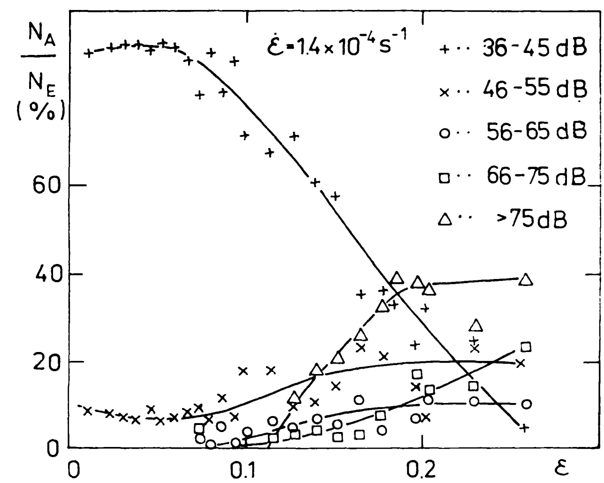

Fig. 4. Strain evolution of the amplitude distribution at $\varepsilon = 1.4 - 10^{-4}$ s^{-1}. N_A is number of events in a class, N_E is total number of events.

of serrations observed at this strain rate (0–9 load drops per 0.001 of strain, the count rate descreased with the decreasing density of serrations). The value of dislocation velocity seems to be very close to the point A in Fig. 1 in this case.

The strain evolution of the amplitude distribution at one strain rate is shown in Fig. 4. A similar behaviour was observed at lower strain rates. At the highest strain rate of the amplitude the distribution was observed not to change (typical values of classes of events N_A/N_E were (35–45) dB – 58 %, (46–55) dB – 26 %, (56–65) dB – 10 %, (66–75) dB – 5 %, more than 75 dB – 1 %). Note some burst activity during the whole experiment at this strain rate.

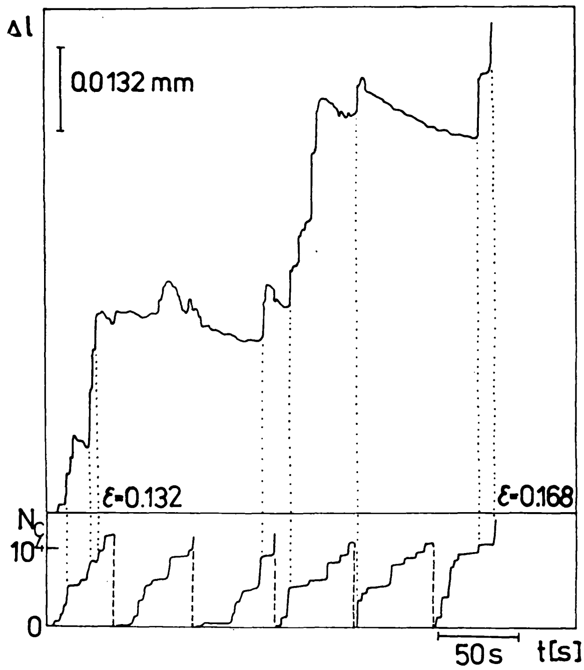

Fig. 5. The part of the extensometer register with corresponding temporal register of total number of counts. The counter was every 50 s read and reset by a computer for obtaining amplitude distributions.

Part of the extensometer register is shown in Fig. 5 together with the corresponding part of total counts register. Some strain steps probably correspond to AE signals as indicated.

4. Discussion

The shape of observed dependences seems to result from a very complex character of the mechanism controlling plastic flow in the Al-3 % Mg alloy.

In case of the lowest strain rate favourable conditions for clustering of excessive solutes in dislocation cores are expected due to a very small value to dislocation velocity (this seems to be supported by resistometric results of Kanadani and Sakakibara [20] and Osamura and Ogura [21]). Small clusters formed in this manner may serve as additional breakable pins and raise the AE activity, in agreement with many papers discussed in [16].

In subsequent stages of deformation or if higher strain rate is applied, other types of obstacles such as dislocation tangles [15] or subboundaries found by Nohara [22] must be taken into account. These obstacles are unbreakable and reduce the AE activity [16] with a possible exception as follows. The amplitude of serrations is known to increase with strain as one can also see in Fig. 2. Thus, growing stress increments are necessary for the first acceleration of every deformation band. This may result in an increase of the density of moving dislocations and large AE bursts should appear. Indeed, a growing contribution of high energy bursts to measured AE with strain is seen in Fig. 4. This was reported also by Cáceres and Rodriguez [17].

In case of the highest strain rate there is probably a lack of obstacles able to decelerate moving dislocations to turn them to the state of saturated core at the onset of the plastic deformation. Hence an alternative explanation of the strain rate dependence of strain at which the PLC effect starts may be proposed to the conventional vacancy model of enhancing bulk diffusivity, namely, that the PLC effect can start only after creating new obstacles by homogeneous deformation. Thus the onset of the PLC effect would shift to a higher value of strain at higher strain rates and the first peak of AE would correspond to ordinary plastic deformation. This situation indeed appeared in our measurements as it is visible in Fig. 3.

5. Conclusion

The results presented seem to support a certain role of the diffusion, especially the core diffusion in the PLC effect. Several types of immobile obstacles, such as dislocation tangles or subboundaries might play an additional role in the mechanism controlling the serrated yielding in subsequent stages of deformation or if higher strain rate ($\dot{\varepsilon} \gtrsim 10^{-3}$ s^{-1}) is applied.

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