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Abstract. The temperature-strain rate domain of the Portevin-Le Chatelier (PLC) effect has been determined for three ultrafine grained aluminum alloys: Al-3Mg (A1), Al-4,57Mg-0,35Mn-0,2Sc-0,09Zr (A2) and Al-5,4Mg-0,52Mn-0,1Zr (A3) (wt.%). The apparent activation energy of the serrated yielding was estimated. It was found that the presence of dispersed particles leads to a decrease in the activation energy and the temperature–strain rate domain of the PLC effect becomes narrower.

1. Introduction

Non-heat treatable Al-Mg alloys are widely used in various industries, in particular, in aircraft, automotive and shipbuilding industries, and operated at cryogenic temperature. Excellent combination of corrosion resistance, weldability, high strength, and good ductility is the main reason for their wide use [1-2].

For a long time, Al-Mg alloys have been used as the common materials for studying the Portevin-Le Chatelier (PLC) effect [3-4]. These alloys are characterized by the instability of plastic deformation in a wide temperature/strain rate interval [4-6]. The PLC effect in Al-Mg alloys can be observed at the room temperature in a certain strain-strain rate interval [7-8] and it is characterized by strain localization in the form of narrow bands of intense shear.

The studying of the conditions for the appearance of the PLC effect is important to prevent its developing during the thermomechanical processing of Al-Mg alloys [9]. However, there is a lack of information on the effect of dispersed particles of the second phase on the PLC effect in aluminum alloys [3]. The purpose of the present study is to establish the influence of the phase composition on the region of existence of the PLC effect in Al-Mg alloys.

2. Materials and method

To reveal the influence of phase composition on PLC effect the following aluminum alloys were investigated: alloy A1: Al-3Mg (wt.%); alloy A2: Al-4.57Mg-0.35Mn-0.2Sc-0.09Zr (wt.%); alloy A3: Al-5.4Mg-0.52Mn-0.1Zr (wt.%). Alloy A1 was prepared by semi-continuous casting. The ingots were further homogenized at 500°C for 4 h, followed by slow cooling in the furnace. Then, billets with a cross-section of 20×20 mm^2 and a length of 110 mm were subjected to a two-stage ECAP at a temperature of 300°C to a strain of ~ 8 and at a temperature of 200°C to a strain of ~ 4. Alloys A2 and A3 were produced by continuous casting. The ingots of the alloy A2 were subjected to homogenization annealing at 360–380°C for 12 h. For the formation of a fine-grained microstructure, ECAP was carried out at 300°C to a strain of ~ 12. The obtained ingots of alloy A3 were subjected to

homogenization annealing at 360°C for 6 h. The ECAP of the A3 alloy was carried out at 300°C to a strain of ~ 12.

The microstructural investigations and elemental analysis were performed by transmission electron microscopy (TEM) by using a JEOL JEM-2100 microscope equipped with an INCA energy dispersive x-ray spectrometer. The flat «dog-bone» shaped specimens with a 16 mm gauge length and 3×1.5 mm2 cross-section were cut parallel to the last extrusion axis of the pressed billets. The uniaxial tensile tests were carried out by using universal testing machine Instron 5882 in the air at temperatures from 398 K to 173 K and strain rates ranging from 1.0×10^{-5} s^{-1} to 2.1×10^{-5} s^{-1}.

3. Results and discussion

3.1. Microstructure after ECAP processing

The bimodal microstructure was observed in alloy A1 after ECAP. Mean size of coarse and fine grains in this condition is 20 and 1.4 μm, respectively. While, ECAP leading to the formation of the uniform fine-grained microstructure with a mean grain size of 1.1 and 1.5 μm in alloys A2 and A3 respectively.

TEM observation did not reveal any precipitates in alloy A1 (figure 1(a)). Whereas it was found that the particles of Al_6Mn and $Al_3(Sc,Zr)$ distributed homogeneously throughout the aluminum matrix in alloy A2 (figure 1 (b)). It is known that the fine particles of $Al_3(Sc,Zr)$ precipitated in aluminum alloys have a coherent interface. The mean size of these particles in alloy A2 is about 10 nm (figure 1b). The mean size of Al_6Mn particles (~80 nm) is rather larger than those of $Al_3(Sc,Zr)$. Only Al_6Mn particles with a mean size of 22 nm were found in alloy A3 (figure 1 (c)).

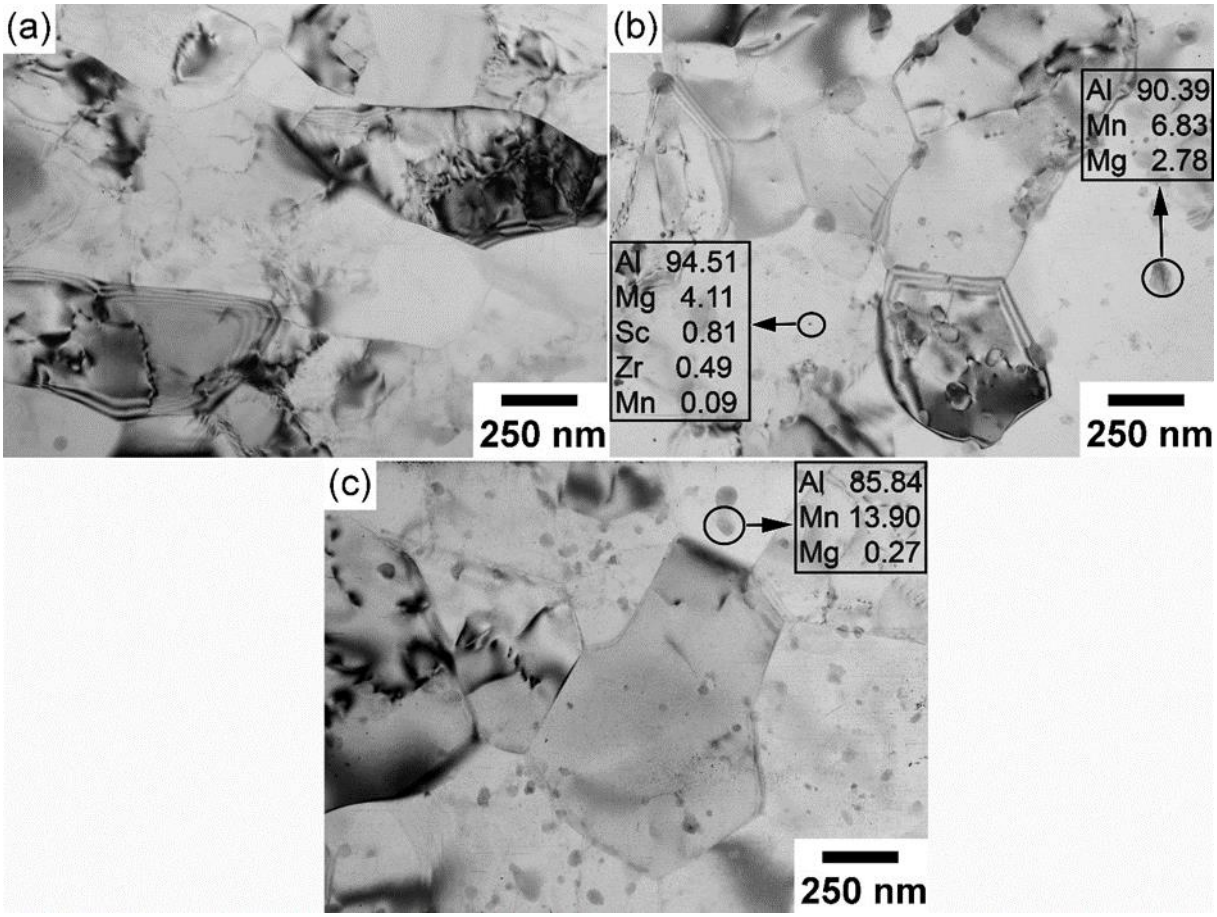

Figure 1. Microstructure of studied aluminum alloys after ECAP processing: A1 (a), A2 (b) and A3 (c). The chemical compositions are given in wt.%.

The presence of dispersed second phase particles precipitated during homogenization enhances the refinement of microstructure. Thus, the ECAP processing of alloy A1 leading to formation of relatively coarse grains. It is obvious that the dispersed particles of Al_6Mn determines the rate of grain refinement, mean size of grains and homogeneity of microstructures formed during ECAP processing.

3.2. Strain rate vs temperature domains of serrated flow in the studied Al-Mg alloys

The analysis of stress-strain curves was performed to determine the presence of PLC effect under different strain rate/temperature conditions in studied alloys after ECAP processing. Figure 2 shows the PLC domains, where solid and open symbols indicate the presence and absence of serrations respectively. It is obvious that the temperature interval of PLC domains of studied alloys tends to increase with decreasing the strain rate.

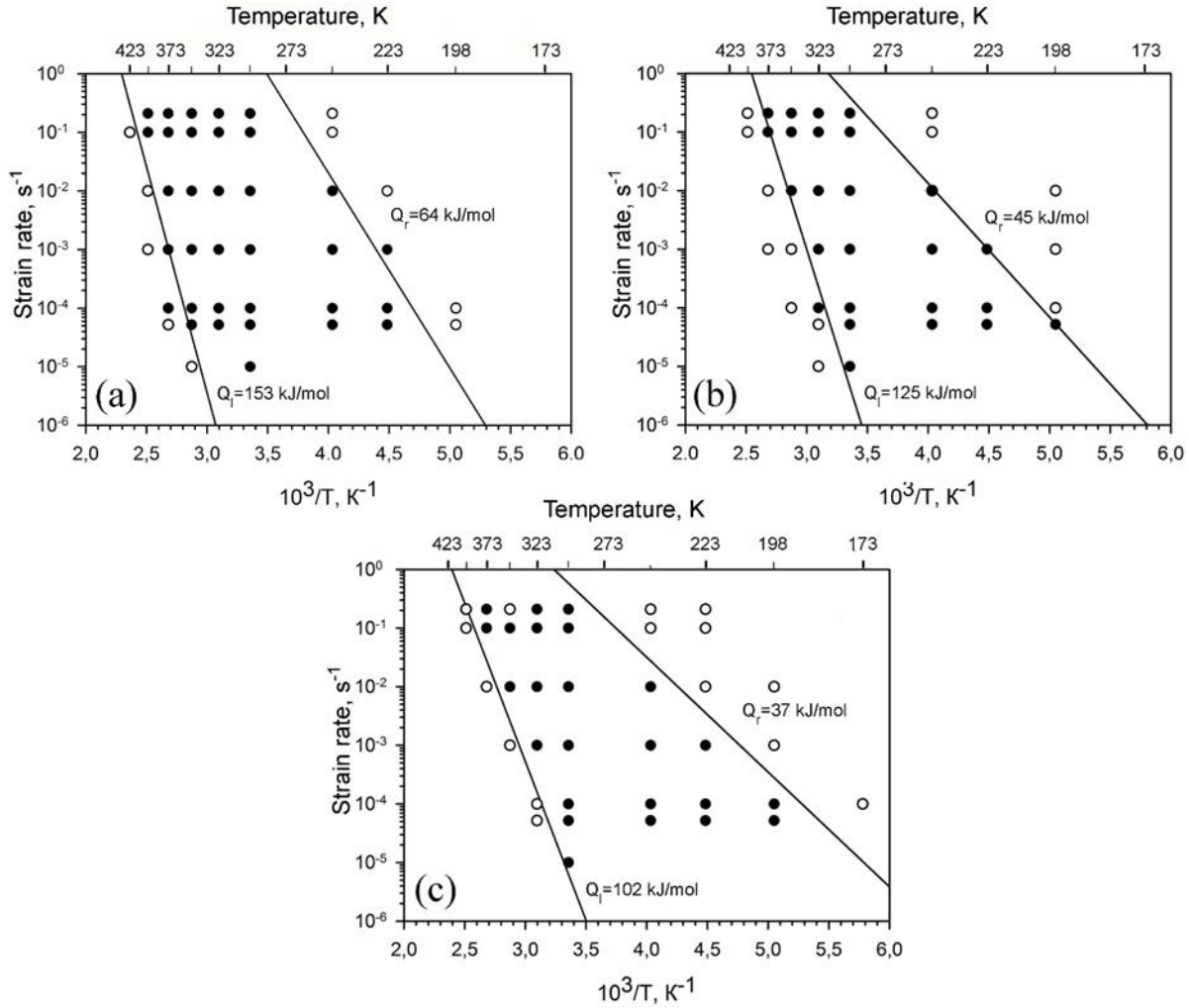

Figure 2. Strain rate vs temperature domains of serrated flow in the Al-Mg alloys: A1 (a); A2 (b) and A3 (c) (open symbols - no serrations, closed symbols - serrations).

In figure (1) it is clearly seen that for each $\dot{\varepsilon}$, the temperature interval of instability extends slightly, as shown by the delimiting boundaries between stable and unstable behavior. Such boundaries can be described by the following equation according to Cottrell theory [10]:

$$\dot{\varepsilon} = \frac{4b\rho C_V D_O}{l}\exp(\frac{-Q}{kT}), \quad (1)$$

where b, ρ, C_v, D_O, l, Q and k represent the Burgers vector, the dislocation density, the vacancy concentration, the diffusion frequency factor, the effective radius of the solute atmosphere, the activation energy for solute migration, and the Boltzmann constant, respectively.

The apparent activation energy of the PLC effect can be evaluated as a slope of the lnέ – 1/T dependencies. The obtained activation energies of the stress serrations of alloy A1 is 64 and 153 kJ/mol for the right (Q_R) and left (Q_L) boundary respectively. It was found that activation energies of stress serrations of alloys A2 an A3 is somewhat lower than that of alloy A1. For alloy A2 the value of QR is 45 kJ/mol while QL is about 125 kJ/mol and for alloy A3 Q_R and Q_L is 37 and 102 kJ/mol respectively. The different values of Q_R and Q_L for the same alloy at high and low temperatures arise from the difference in mechanisms controlling the unstable flow. The transition from type C serrations at high temperatures to type A serrations at low temperatures was observed for alloy A1. In contrast, the mixed type A+B takes place at the strain rate of 1×10^{-3} s^{-1} in alloys A2 and A3 and is replaced by pure type A on the right boundary (figure 2 (b), (c)). The obtained results agree with the present understanding of PLC effect, which involve the influence of different microstructure parameters on the jerky flow behavior [11].

4. Summary and conclusions

The analysis of the stress flow serrations in three aluminum alloys with different phase composition was performed. The following results are found:

1. The difference in the values of apparent activation energies for the right and left boundaries of the PLC domain is due to the transition between different types of stress flow serrations.
2. The dispersed particles in an aluminum matrix lead to a decrease in the apparent activation energy for both boundaries.
3. The temperature interval of PLC domain of alloys A2 and A3 is lower than that of alloy A1. Such difference can be related to the presence of dispersed particles in alloy A1 and A2.

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