Experimental characterization of Portevin-Le Chatelier instabilities in Al-2.5% Mg alloy

CHIBANE Nouara[a]

[a]. Laboratoire de Physico-chimie des Matériaux et Catalyse (LPCMC), Faculté des Sciences de la Nature et de la Vie, Université de Bejaia, 06000 Bejaia, Algérie. Chibane_n@hotmail.fr

Résumé

Le présent travail est consacré à l'étude des instabilités de la déformation plastique de type Portevin-Le Chatelier (PLC). Nous avons analysé les caractéristiques temporelles des instabilités PLC dans l'alliage Al-2.5%Mg à température ambiante. Des essais de traction uniaxiale, effectués dans la gamme des vitesses de déformation allant de $5.10^{-5}s^{-1}$ *à* 10^{-1} s^{-1}*, nous ont permis de déterminer le domaine d'instabilité et d'étudier les évolutions des paramètres associés aux instabilités en fonction de la déformation et de la vitesse de déformation. Les résultats expérimentaux obtenus ont été discutés en liaison avec les mécanismes de vieillissement dynamique.*

Abstract

In the present work, we have investigated the Portevin-Le Chatelier (PLC) effect at room temperature in Al-2.5%Mg alloys. Using tensile tests under constant applied strainrate, ranging from 5.10^{-5} *to* 10^{-1} s^{-1}*, we determined the strain/strain rate domain of instability of plastic flow. The PLC characteristics are analyzed and the obtained results are interpreted in accordance with Dynamic Strain Aging (DSA) models.*

Mots clefs : Portevin-Le Chatelier effect, Localized strain, Dynamic Strain Aging, Ductility

1 Introduction

The instable plastic flow constitutes a major inconvenient during the formability of metallic materials. The Portevin-Le Chatelier (PLC) is one of these plastic heterogeneities [1-4]. It leads to heterogeneous mechanical properties, reduces the ductility of the deformed material and creates areas very sensitive to the corrosion. The strain localization zones are characterized by a dilating behavior which can cause the material rupture and, consequently, the failure of structures [1, 2]. The optimization of the homogeneous material formability is based mainly on the results of characterization and modeling of the unstable plastic flow. PLC instabilities are observed in different materials with different histories and in different conditions. In Al–Mg alloys, which provide a large variety of applications due to their low weight and high mechanical strength, jerky flow appears around room temperature in a limited range of strain and strain rate [1-3]. The PLC effect is characterized by a macroscopic spatio-temporal localization of plasticflow. It appears in the form of a repeated stress drops in the stress-strain curve at imposed strain rate. The produced strain localizations on the sample surface (PLC bands) are static at low strain rates (Type C). They move by jumps at intermediate strain rates (Type B) and propagate at elevated applied strain rates (Type A). Initiation of plastic flow is generally preceded by a certain critical plastic strain ε_c, which is strongly dependent on the temperature and on the applied strain rate [1-3]. The microscopic origin of the PLC effect is associated to the dynamic strain aging (DSA) phenomenon resulting from the interaction between mobile dislocations and the clouds of impurities [1, 4, 5]. The solute atoms diffuse towards dislocations during their temporary arrests at local obstacles and increases, consequently, the plastic flow stress. The repeated breakaway of dislocations from the solute clouds reduces the strain rate sensitivity (SRS) of the flow stress, which becomes negative. Therefore, the strain localizes into narrow deformation bands and gives rise to serrated stress-strain curve at constant applied strain rate.

Through uniaxial tensile tests, along the rolling direction, the amplitude of serrated yielding, the reloading time, the strain rate sensitivity and the critical strain for the onset of jerky flow have been studied and discussed as a function of strain and strain rate at room temperature.

2 Experimental

The material used in the present study is an industrial Al-2.5%Mg alloy. Its chemical composition in weight percent isAl-2.52%Mg-022%Mn-0.37%Cr-0.09%Si-0.35%Fe-0.02%Cu.

Polycrystalline flat samples (gauge length 42 mm, width 6 mm, thickness 2.25 mm, radius of curvature of the shoulders 4 mm). The samples are deformed in tension with an Zwick/Roell hard testing machine, i.e., at constant driving velocity, at room temperature (22.5°C) and at strain rates in the range 5.10^{-5} - 10^{-1} s^{-1}, corresponding, to the PLC range in Al-Mg alloys around the room temperature [1,5,12,14]. All tensile tests were carried out in similar conditions.In what follows; we studied jerky flow in the Al-2.5%Mg alloys during tensile tests at different applied strain rate and at room temperature.

3 Results and discussion

3.1 Deformation curves and PLC instability types

In the investigated strain rate range 5.10^{-5}–10^{-1} s^{-1}, at room temperature, the macroscopic strain rate sensitivity of the flow stress was found to be negative in the studied Al–2.5%Mg alloys. Indeed, for a given plastic strain, the higher level of jerky flow decreases with increasing applied strain rate. PLC characteristics depend on the imposed strain rate, strain hardening and the material state. In both types of samples, PLC instabilities shift from the type A to the type B then to the type C with decreasing of the imposed strain rate. In fact, when one reduces the strain rate, the waiting time of dislocations at local obstacles increases and, consequently, the DSA becomes more intense and the heterogeneous plastic flow more accentuated [1, 8, 9,11]. Fig.1 shows the serrated stress-strain curves in the Al-2.5% Mg alloy obtained at room temperature for different imposed strain rates. Three types of instabilities are observed. At high strain rates, $\dot{\varepsilon}=10^{-2}$ s^{-1} and $\dot{\varepsilon}=10^{-3} s^{-1}$(Fig.1-a and Fig.1-b), instabilities are of Type A. They are in the form of stress drops, corresponding to bands nucleation, separated by weak undulations dues to their continuous displacements. Type A Bands generally initiate at the end of the sample and propagate continuously until the opposite one [2].

At intermediate strain rates, $\dot{\varepsilon}=10^{-4}$ s^{-1}, the stress drops are abrupt and are characterized by elastic reloading between two successive drops (instabilities called of Type B). They are usually structured in stages separated by large drops, as shown in the insert of Fig.1-c. Every high stress drop corresponds to the PLC band appearance, which moves by successive jumps producing additional stress drops of lower magnitude. Thus, each PLC band initiates after the other giving rise to an apparent displacement (correlated hopping band).

At very low strain rates, $\dot{\varepsilon}=10^{-5}$ s^{-1}, instabilities are characterized by very large amplitudes and, partially, by a plastic reloading between two successive drops (Fig.1-d). These instabilities are called of Type C. The associated strains localizations occur randomly on the deformed sample.

The magnitude of the stress drops is one of the most conspicuous characteristics of the PLC instabilities. It is highly linked to the importance of DSA and then it reveals the heterogeneity grade of plastic flow [1, 2]. The deformation becomes more heterogeneous as the magnitude of the stress drops is important. Indeed, for an imposed strain rate, the obstacles density (forest dislocations) increases when the strain increases and, therefore, the average waiting time of mobile dislocations increases. Thus, the time available to solute atoms diffusion is enhanced and, therefore, the DSA becomes more and more significant and the magnitude of the stress drops increasingly large. When the applied strain rate increases, at a given strain, the waiting time decreases and the DSA becomes less significant producing low stress drops in the stress-strain curve.

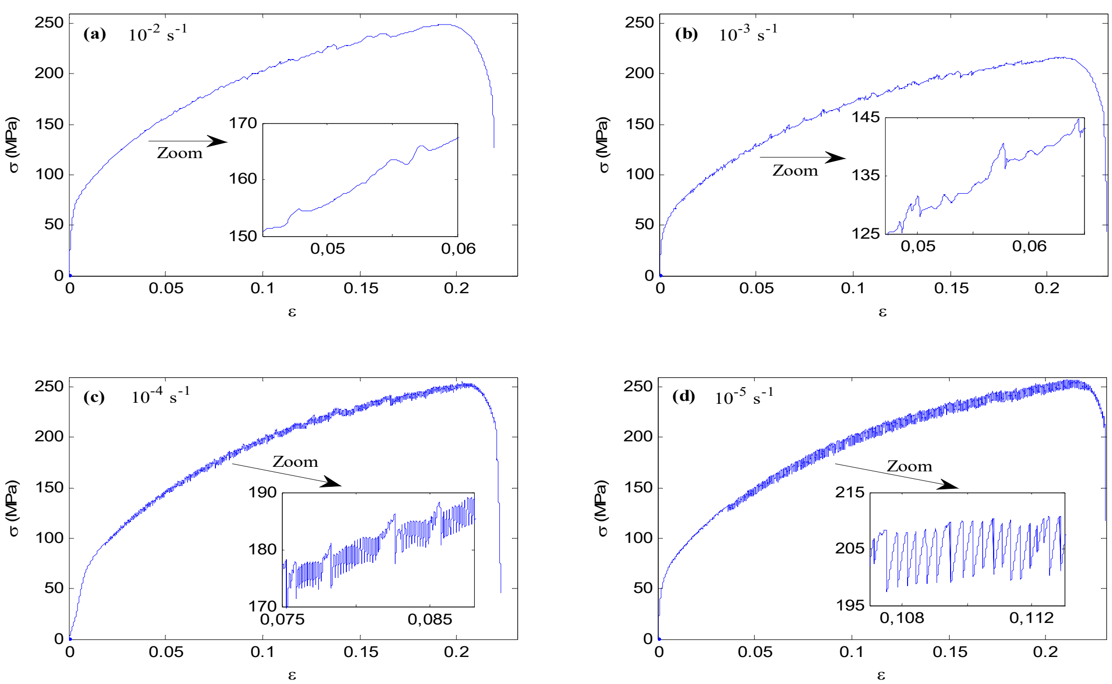

Fig. 1 –Stress-strain curves at different applied strain rates of the Al-2.5%Mg alloy at room temperature. (a) 10^{-2} s^{-1}, (b) 10^{-3} s^{-1}, (c) 10^{-4} s^{-1}, (d) 10^{-5} s^{-1}.

3.2 Domain of PLC instability

For a given strain rate, plastic flow becomes unstable beyond a certain plastic critical strain ε_c and remains as such until failure by necking of the sample. According to DSA models [6, 9, 11, 12], ε_c corresponds to the necessary strain for appearance of negative SRS. It depends on the material structure and history, the applied strain rate as well as the temperature. The domain of instability and the strain rate dependency of the plastic critical strain for onset of discontinuous yielding ε_c and the plastic strain at failure ε_r are shown in Fig.2. The plastic critical strain ε_c decreases then increases with increasing applied strain rate, behaviors known as “inverse” and “normal”, respectively. At high strain rates, ε_c presents a normal behavior, where it increases when the applied strain rate increases, and an inverse behavior at low strain rates where it decreases with strain rate.

DSA models [4, 5] explain the normal behavior of the critical strain. It is related to the vacancy generation and dislocations multiplication during plastic deformation. On the other hand, the inverse behavior is still the subject of controversies of many models [6-8]. The combined action of DSA and the presence of sheared precipitates by dislocations can be the cause of this behavior. The strain rate dependency of the plastic strain at failure ε_r (Fig.2) shows a ductility reduction of the Al-2.5% Mg alloy at room temperature. In fact, ε_r decreases then increases when the applied strain rate increases. The ductility reduction is a consequence of the SRS reduction in the strain rate PLC domain.

With increasing strain, for an imposed strain rate, the observed change on PLC instability type (from type A to type B and from type B to type C) and the increase in the magnitude of the stress drops indicate that DSA becomes more and more significant. Indeed, as strain increases the obstacles to dislocation motion become more and more difficult to overcome. Consequently, by increasing the average waiting time of mobile dislocations at forest obstacles, effect of DSA is enhanced and the heterogeneity of the plastic flow becomes more and more accentuated. This is translated on the stress–strain curves by transitions between the types of instability.

The minimum of the curve $\varepsilon_c(\dot{\varepsilon}_a)$, which corresponds to the switch from “inverse” to “normal” behavior, is situated around 2.10^{-3} s^{-1} in the present alloy. It correlates with a gradual shifting of PLC bands from type A to type B with increasing strain, in agreement with results previously reported in Al–Mg alloys [2, 7, 9]. The plastic strain at failure ε_r first decreases, then increases when the applied strain rate increases. Hence, a reduction of the ductility of the Al-2.5%Mg alloy occurs at room temperature in the PLC domain. The parallel evolutions of ε_c and of ε_r suggest that the reduction of the ductility is a direct consequence of the negative SRS [1, 10].

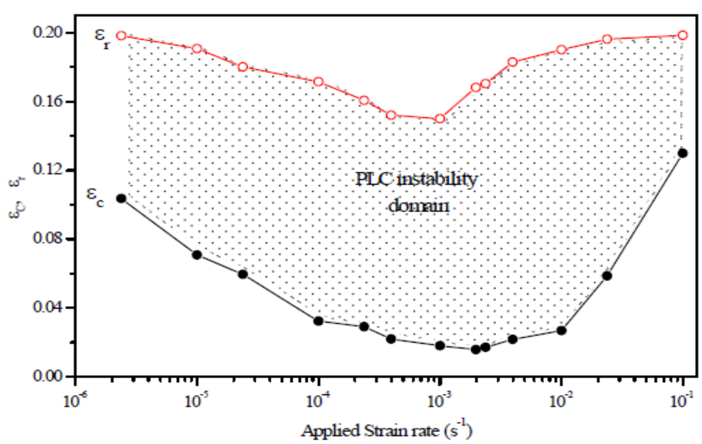

Fig. 2 –PLC instability domain of the Al-2.5%Mg alloy at room temperature. Effect of the imposed strain rate on the critical plastic strain ε_c and the plastic strain at failure ε_r .

3.2 Amplitude of serrations and the reloading time between two successive stress drops

The magnitude of the stress drops is one of the most conspicuous characteristics of the PLC instabilities. It is highly linked to the importance of DSA and then it reveals the heterogeneity grade of plastic flow [1, 2]. The deformation becomes more heterogeneous as the magnitude of the stress drops is important.

Fig.3 (a) and (b) shows that the magnitude of serrations and the reloading time between two successive stress drops increase with strain and decrease with strain rate. The average SRS, calculated between two successive applied strain rates, is negative and decreases with increasing strain in all tensile tests. This means that the DSA intensifies during straining. The increase in the obstacles density and, consequently, the waiting time of mobile dislocation becomes significant and leads to an increase in dislocation pinning by the diffusion of the solute atoms. These results are in agreement with several earlier reports [2]

(a) (b)

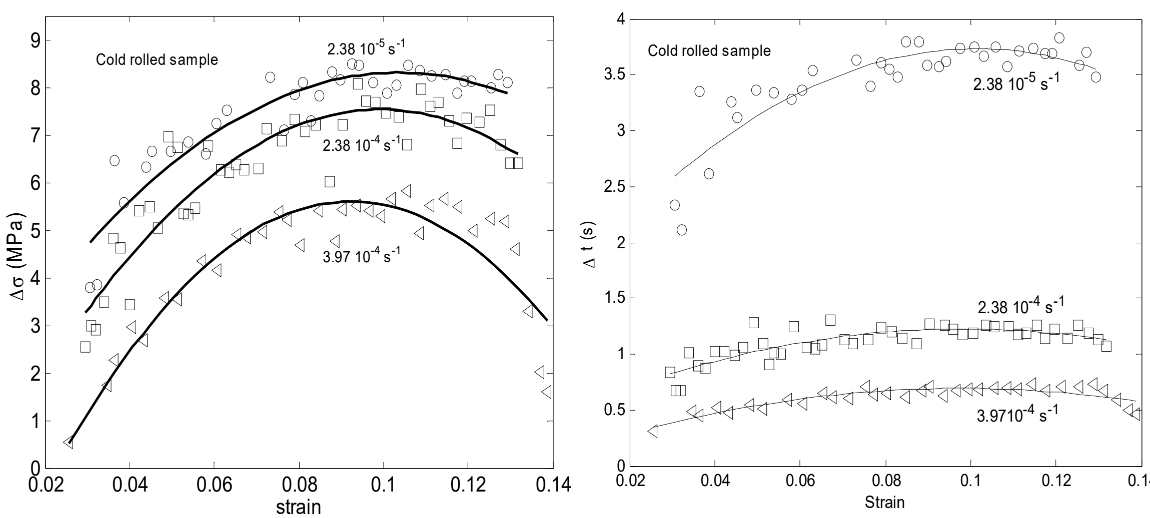

Fig. 3. (a) The magnitude of serrations between two successive stress drops; (b) The reloading time between two successive stress drops.

4 Conclusion

The Portevin-Le Chatelier (PLC) effect is investigated at room temperature in Al-2.5%Mg alloys. Using tensile tests under constant applied strain rate, the PLC characteristics are analyzed. The study shows that, for a given strain, instability shifts from Type C to Type B then to Type A when the applied strain rate is increased. With increasing strain, for an imposed strain rate, the observed change on PLC instability type (from type A to type B and from type B to type C) and the increase in the magnitude of the stress drops indicate that DSA becomes more and more significant. Indeed, as strain increases the obstacles to dislocation motion become more and more difficult to overcome. Consequently, by increasing the average waiting time of mobile dislocations at forest obstacles, effect of DSA is enhanced and the heterogeneity of the plastic flow becomes more and more accentuated.

This is translated on the stress–strain curves by transitions between the types of instability. The necessary critical plastic strain for the onset of PLC instabilities manifests a normal behavior at high strain rates, where instabilities are of Type A, and an inverse behavior at low strain rates, where the instabilities are of Type B or Type C. The strain rate dependency of the plastic strain at failure shows a ductility reduction of the Al-2.5% Mg alloy in the strain rate PLC domain at room temperature. The amplitude of serrations and the reloading time between two successive stress drops increase with strain and decrease with strain rate. The average SRS, calculated between two successive applied strain rates, is negative and decreases with increasing strain in all tensile tests. This means that the DSA intensifies during straining.

References

[1] Kubin L.P., Estrin Y., Dynamic strain ageing and the mechanical response of alloys, J. Phys III, 1, (1991) 929-943.

[2] Ait-Amokhtar H., Boudrahem S., Fressengeas C., Spatiotemporal aspects of jerky flow in Al–Mg alloys,in relation with the Mg content, Scripta Mater, 54, (2006) 2113-2118.

[3] Ziani L., Boudrahem S., Ait-Amokhtar H., Mehenni M., Kedjar B., Unstable plastic flow In the Al–2%Mg alloy, effect of annealing process, Mater. Sci. Eng.A, 536, (2012) 239- 243.

[4] McCormick P.G., A model for the Portevin Le Chatelier effect in substitutional alloys, ActaMetall, 20, (1972) 351-354.

[5] Van Den Beukel A., Theory of the effect of dynamic strain aging on mechanical properties, Phys Status Solidi A, 30, (1975) 197-206.

[6] Kumar S., Inverse behaviour of the onset strain of serrated flow, ScriptaMetall.Mater, 33, (1995) 81-86.

[7] Hahner P., on the critical conditions of the Portevin Le Chatelier effect, Acta Mater, 45, (1997) 3695-3707.

[8] Brechet Y., Estrin Y., on the influence of precipitation on the Portevin-Le Chatelier effect, ActaMetall.Mater, 43, (1995) 955-963.

[9] L.P. Kubin, C. Fressengeas, G. Ananthakrishna, Collective Behaviour of dislocations in Plasticity, in : F.R.N. Nabarro, M.S. Duesbery (Eds), Dislocations in Solids, Vol. 11, Elsevier Science BV, (2002) 101-192.

[10] H.Ait-Amokhtar, S. Boudrahem, C.Fressengeas, Scripta Mater.54, (2006) 2113-2118.

[11] A. Van den Beukel, Phys. Stat. Sol. (a) 30, (1975) 197 - 206.

[12] P. Peninng, Acta Metall., Vol. 20, (1972).

[13] R.Kral, P.Lukac, Mater .Sci.Eng. A234 - 236, (1997) 786 - 789. .

[14] H. Ait-Amokhtar, C. Fressengeas, Acta Mater 58, (2010)1342 - 1349.