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On the serrated yielding in AlMg6.5 alloy sheet

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Room temperature tensile tests were performed on AlMg6.5 alloy sheet, with initial strain rates of 6.7×10^{-4} s^{-1}, 6.7×10^{-3} s^{-1} and 6.7×10^{-2} s^{-1}, after imposing 70% cold rolling reductions, and annealing at 320 °C/3h. The Portevin-Le Chatelier effect is the main mechanism of unstable plastic flow, caused by Mg solute-dislocation interactions. Although, a comprehensive comparison of the energies of activation, relevant for serrated yielding, revealed a wide range of values, calculations based on the used strain rates, retained vacancy concentration and the Mg solute diffusion coefficient at room temperature, indicate that serrations are expected to appear from the start of plastic deformation. In other words, under the performed test conditions, the Mg solute atoms are mobile enough for the majority of dislocations to travel in conjunction with the Mg atmospheres. After deformation at a strain rate of 6.7×10^{-2} s^{-1} the "*D*" type serration observed, as a series of successive not completely flat stress plateaus, indicates a complex deformation process, affected by strain hardening, coused by dislocation reactions and the propagation of a kind of non-hardening deformation bands.

Key words: AlMg alloy, serrated yielding, diffusion, activation energy, deformation behaviour.

The flow stress serrations and related surface markings, as a generally occuring phenomena in Al-Mg alloys,[1] are of concern in industrial forming processes. As well as the negative strain rate sensitivity (SRS) associated with its occurrence, tends to limit the tensile ductility by inducing a localized shear failure.[2,3] A main goal would be the ability to predict the strain rate and temperature regimes outside of which serrated yielding does not occur. In this rescept, the strain rate dependence of critical strains, recently examined at room temperature in a Al3Mg alloy,[4] which shows good agreement with the prediction of the model of Kubin and Estrin,[5] seems to be promising. Further, the appearance of "*D*" type serrations as a series of successive plateaus of constant flow stress, under certain conditions, seems to be an important manifestation of some specific hardening effects during plastic deformation.[6,7]

The aim of the present work was to consider the room temperature appearance of serrated yielding, and the observed strain bursts, relevant to hardening in the tested AlMg6.5 type sheet metal. Also, a comprehensive analysis was made in

respect to the discrepancies between the values of the activation energy for serrated yielding in Al-Mg alloys.

EXPERIMENTAL

Material

The as received AlMg6.5 type sheet (2.5 mm thick), with the chemical composition in wt%, of Al-6.5Mg-0.64Mn-0.1Si-0.2Fe-0.03Zn-0.05Ti, was cold rolled 70%, and annealed at 320 °C/3h, in an inert gas atmosphere.

Tensile testing

Room temperature tension test were carried out using small ASTM tension specimens with a 25 mm gauge length. In continuous tests, three cross-head rates were applied: 1mm/min, 10mm/min and 100 mm/min, giving initial strain rates of $\dot{\varepsilon}_1 = 6.7\times10^{-4}\ s^{-1}$, $\dot{\varepsilon}_2 = 6.7\times10^{-3}\ s^{-1}$ and $\dot{\varepsilon}_3 = 6.7\times10^{-2}\ s^{-1}$, respectively.

RESULTS

Discontinuous yielding was observed in the tested AlMg6.5 alloy within a limited interval of strain rates, $\dot{\varepsilon}$, ranging from $6.7\times10^{-4}\ s^{-1}$ to $6.7\times10^{-2}\ s^{-1}$. The flow curves do not exhibit critical strains, separating regions of smooth and jerky flow, *i.e.*, serrated yielding starts from the beginning of plastic deformation (Fig. 1). At the strain rate of $6.7\times10^{-4}\ s^{-1}$ (curve 1), the serrations are of a form such that the load falls below the general level of the curves, *i.e.*, they exhibit type *C* serrations, which increase gradually as the deformation proceeds.

On increasing the strain rate to $6.7\times10^{-3}\ s^{-1}$ this alloy exhibits type *A* serrations or combinations of *A* and *B* serrations (curve 2 in Fig. 1). At the strain

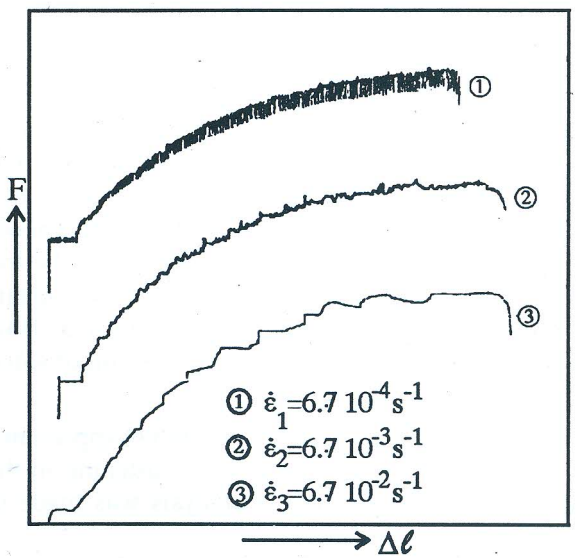

Fig. 1. Load *vs.* extension curves at different strain rates.

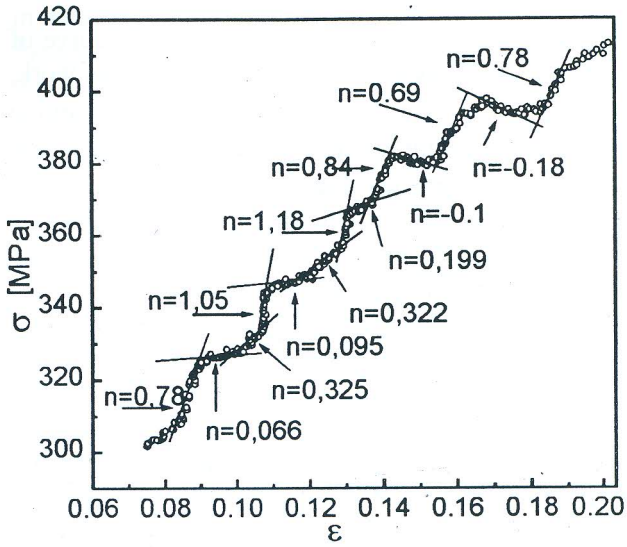

Fig. 2 Enlarged part of curve 3 in Fig. 1, showing successive stress jumps and plateaus with defined slopes, after straining at a true strain rate of 6.7×10^{-2} s^{-1}.

rate of 6.7×10^{-2} s^{-1} serrations were observed to occur in a series of successive plateaus of constant stress, separated by stress jumps $\Delta\sigma$ (curve 3 in Fig. 1), sometimes described as *D* or combiined *D*+*B* serrations.[8] The length of these plateaus, $\Delta\varepsilon_b$, increases during deformation. It was found by fitting the general shape of curve 3 that the strain hardening exponent is $n = 0.35$. On the other hand, the plateaus have slopes of varying values, including positive and negative values (see Fig. 2), giving an average $n_{av} = 0.46$.

DISCUSSION

On the origin of serrated yielding at the start of plastic deformation

It is obvious from Fig. 1 that serrated yielding occurs from the very beginning of plastic deformation. At the strain rate of 6.7×10^{-4} s^{-1}, the dislocations can easily be locked by Mg solute atmospheres, but at higher strain rates the velocity of the mobile dislocations is high enough to break away from their atmospheres. However, the Mg solute mobility is also high enough to catch up and move with the mobile dislocations. Since the critical temperature, T_c, above which Portevin - Le Chatelier effect (*PL*) occurs is ≈ 0 °C[3,9] for Al-Mg based alloys, the equilibrium vacancy concentration and the Mg solute mobility are large enough to allow serrated yielding to start occuring from the beginning of plastic deformation. Furthermore, the Mg solute mobility can be enhanced by the increasing vacancy concentration introduced by the deformation.

According to the procedure used by King *et al.*[3] and the estimation by Cottrell[10] that serration takes place in aluminum alloys when:

$$D \approx 10^{-14}\,\dot{\varepsilon}\ [\mathrm{m^2\ s^{-1}}] \tag{1}$$

it is possible to estimate the appearance of serrations on the stress-strain curve of the investigated Al-6.5 % Mg alloy. For the slowest strain rate used in this work, 6.7×10^{-4} s^{-1} Eq. (1) gives $D \approx 6.7\times10^{-18}$ m^2 s^{-1}. Taking: $D_0 \approx 10^{-4}$ m^2 s^{-1} and $Q \approx 131$ kJ mol^{-1},[11] or ≈ 150 kJ mol^{-1},[3] the room temperature diffusion coefficient for Mg (present in a large quantity) can be calculated to be in the range from $D \approx 4.3\times10^{-28}$ m^2 s^{-1} to $D \approx 1.8\times10^{-31}$ m^2 s^{-1}, respectively. It is concluded from this calculation that serrated yielding would not be expected, due to the activation energy gap, until plastic deformation had increased the vacancy concentration.

According to Cottrell,[12] the deformation induced vacancies will enhance the diffusion coefficient by:

$$D \approx 0.12\ C_v \exp\left(-E_{mV}/RT\right) \tag{2}$$

where E_{mV} is the energy of activation for the exchange of vacancies and solute atoms, and C_v is the vacancy concentration. E_{mV} is a part of the activation energy for the diffusion of a substitutional solute. For this reason E_m in the Cottrell[13] equation:

$$\varepsilon_c^{\,m} = A\times\dot{\varepsilon}\times\exp\left(E_m/RT\right) \tag{3}$$

refers to the energy of activation for the migration of vacancies rather than solute. According to Seitz,[14] this vacancy fraction increases with strain:

$$C_v = B\times\varepsilon\times\exp\left(m\right) \tag{4}$$

where ε is the strain, and B and m are constants. The value of the exponent in this vacancy concentration relation is usually between 1 and 2 for substitutional alloys and 0 for interstitial alloys. Based on this consideration, it is to be expected that for each strain rate and temperature there will be a critical strain, ε_c, as which serration begins.[3] After the specimen is quenched from ≈ 320 °C $D \propto 6\times10^{-18}$ m^2 s^{-1}. For the range of strain rates used in our experiments, 6.7×10^{-4} s^{-1} to 6.7×10^{-2} s^{-1}, Eq. (2) predicts that for serrations to occur the diffusion coefficients need to be:

$D_1 \approx 6.7\times10^{-18}$ m^2 s^{-1} for $\dot{\varepsilon}_1 = 6.7\times10^{-4}$ s^{-1}

$D_2 \approx 6.7\times10^{-17}$ m^2 s^{-1} for $\dot{\varepsilon}_2 = 6.7\times10^{-3}$ s^{-1}

$D_3 \approx 6.7\times10^{-16}$ m^2 s^{-1} for $\dot{\varepsilon}_3 = 6.7\times10^{-2}$ s^{-1}

It is obvious that all D values lie below $D \approx 6\times10^{-18}$ m^2 s^{-1}, *e.g.*, it is expected that serrated yielding will start with zero plastic strain (theoretically) or with a minimum plastic strain (experimentally),[11] as was observed in our experiments (see Fig. 1). This model, based on vacancy assisted dynamic strain aging, is only valid if the vacancy concentration present at 320 °C is retained on accelerated cooling to room temperature, which is a resonable assumption because of the high binding

energy in the range[15] of 11.6 and 27.1 kJ mol^{-1},[16] between Mg atoms and vacancies in an aluminum matrix. The high Mg atom/vacancy binding energy in aluminum, E_{bMgV}, is caused by a large relaxation of elastic-strain energy associated with placing a Mg atom adjacent to an Al matrix atom in a perfect lattice, the origin of which lies in the large difference (11.7%) in diameters of Al and Mg atoms. According to Faulcner,[17] the electronic contribution arising from interaction between outer-electron orbital of the Mg and the Al matrix atoms, E_e, is much smaller, *i.e.*, the relaxation term, E_r, in equation:[18]

$$E_{bMgV} = E_e + E_r \tag{5}$$

has the dominant effect.

The calculation of Miura[19] shows that the energy of activation for the onset of serrated yielding, E_{aSY}, in Al-Mg alloys, 57.8 ± 2 kJ mol^{-1} (generally, depending on the serration type, it is in the range between 44 and 70 kJ mol^{-1} [20]) is substantially larger than the energy of activation E_{mMg} for the diffusion of Mg in Al, which is in the range from 33.7 kJ mol^{-1},[21,22] to 53 kJ mol^{-1},[23-25] or, as a Q value, it is in the range from 131 kJ mol^{-1} to 150 kJ mol^{-1},[3,11] because Q/E_m can be as high as 2.5 to 3 for an alloy in equilibrium conditions.[11,26]

The difference between Miura's[19] value of 57.8±2 kJ mol^{-1} and the values for the energies of activation E_{mMg} for the diffusion of Mg in Al, which are in the range from 33.7 kJ mol^{-1},[21,22] to 53 kJ mol^{-1},[23-25] can be rationalized by the presence of a significant number of vacancies in the Al matrix before the start of the test. As a result of the high binding energy, E_{bMgV}, between Mg and vacancies, which has been determined as 11.6 kJ mol^{-1},[15] 24 kJ mol^{-1},[8] and 27 kJ mol^{-1},[20,21] and the high Mg solute content in the investigated alloy, there is reason to believe that the Mg solute will trap 1.2×10^{-6} vacancies, present at the annealing temperature of 320 °C. In such a case, assuming the trapped vacancies are already present in the matrix, one can obtain:

$$E_{aSY} = (E_{fV} - E_{mV}) - E_{bMgV} \tag{6}$$

where E_{fV} and E_{mV} are energies of the activation for formation and movement of vacancies in aluminum, respectively. Assuming a sufficient number of quenched-in vacancies present at the start of the deformation, *i.e.*, $E_{fV} = 0$, and substituting typical values in Eq. (6), one obtaines:

$$E_{mV} = E_{aSY} - E_{bMgV} = 57.8 - 24.06 = 33.74 \text{ kJ mol}^{-1} \tag{7}$$

This value lies between the substantially larger value of 58 kJ mol^{-1} derived by Charnock[26] and the smaller values of 19.3 kJ mol^{-1} and 24.13 kJ mol^{-1} obtained by Harris[27] in an Al-7 wt.% Mg and Al-5 wt.% Mg alloy, respectively. King *et al.*[3] obtained E_{aSY} = 19.7 kJ mol^{-1} in Al-Zn-Mg base alloys The reasons for the large deviations of E_{aSY} values present in the literature may be due to:

a) Cottrell's expressions only hold at low temperatures, around $0.3T_m$, depending on the strain rate and Mg content, as was suggested by Charnock.[26]

b) Instead the interaction of mobile Mg solute with the moving dislocations, the Mg solute atoms diffuse along dislocations at forest intersections, as was suggested by Mulford and Kocks.[28]

c) The number of quenched-in vacancies, present at the start of deformation, is not sufficient to cause the E_{fV} term to neglected, as has been suggested by King *et al.*,[3] and some other authors.

This value is comparable with the ragne of energy of activation values 35 to 55 kJ mol^{-1} for serrated yielding in Al-Mg base alloys.[21,29-34] Similar discrepancies have been observed for lithium in Al-Li base alloys.[35] The vacancy concentration is expected to be sufficiently high at room temperature to allow serrated yielding to start at zero critical strain, *i.e.*, $\varepsilon_c = 0$, which is in very good agreement with Charnock's[26] explanation of the effect of critical temperature ($T_c \approx 0$ °C) for serrated yielding in the behavior of Al-5 wt.% Mg, and other substitutional alloys.

TABLE I. The values of energies of activation: E_{mV}, E_{aSY}, and E_{mMg}, relevant for serrated yielding processes in Al-Mg based alloys [kJ mol^{-1}]

Alloy	E_{mV}**	E_{aSY}	E_{mMg}	References
Al-Mg	72.4[61]		53.1	[36]
Al-3% Mg	24.1	62.7		[27]
Al-7% Mg	19.3	57.9		
Al-3% Mg	72.4[61]	48.3	53.1	[23]
		range: 40.5-54.1		
Al-5% Mg	33.8	96.5		[22]
Al-Mg			118-160	[3]
Al-5.6% Mg	72.4[61]		53.1	[16]
Al-08.% Mg	72.4[61]		53.1	[24]
Al-0.7% Mg	72.4[61]		53.1	[25]
Al-5% Mg		44-55 for *A* (50-65)*		[20]
		55-62 for *B* (≈40)*		
Al-Mg		57		[19]
Al-Mg		26.1-43.4		[13]
Al-5.6% Mg	26.1			[16]
Al-8.6% Mg				
Al-3% Mg		32.8		[37]

TABLE I. Cont

Alloy	E_{mV}**	E_{aSY}	E_{mMg}	References
Al-1% Mg				
Al-3% Mg	33.8	82		[38,39]
Al-5% Mg				
Al-3% Mg	53.1[66,67]	26.1-45.4		[40]
Al-1% Mg				

E_{bMgV} - Binding energy between a vacancy and a Mg atom in aluminum, which is in the range from 19.3 kJ mol^{-1},[41] to 27.1 kJ mol^{-1}.[16]
E_{bDMg} - Binding energy between dislocations and Mg atoms, estimated as being between 19.3 kJ mol^{-1},[13] and 27.1 kJ mol^{-1},[16] or even 48.3 kJ mol^{-1},[9]
E_{mV} - Energy of activation to move a vacancy in aluminum
E_{aSY} - Energy of activation for serrated yielding
E_{mMg} - Energy of activation for Mg diffusion in Al
*The values in brackets were determined by the method developed by Pink and Grinberg[42]

The energy of activation for serrated yielding, E_{aSY}, is closely related and may be compared to the binding energy between the Mg solute and the dislocations, E_{bMgD}. However, it should be noted that the available data in the literature for the binding energy between a vacancy and Mg solute E_{bMgV},[30,40] the activation energy for Mg solute diffusion in Al, E_{mMg},[3,23-26] as well as the activation energy for serrated yielding of Al-Mg type alloys, E_{aSY},[19,20,26,29-35,43] vary considerably. So, it is to be expected that considerations, regarding serrated yielding in these alloys, based on the Mg solute/vacancy binding energy,[44] and/or the binding energy between the Mg solute and dislocations,[20] may not be sufficiently conclusive.

The deformation behaviour

The decrease in the magnitude of the load diminishing with increasing strain rate can be rationalized by a decrease in the available time for the formation of a Mg solute atmosphere around the mobile dislocations. Increasing strain rate, ε, would result in a decrease of the waiting time of the mobile dislocations at obstacles and, consequently, would make the Mg solute atmospheres formed around waiting dislocations more dilute. Such strain aged dislocations require a smaller stress in order to be released and continue to move.

In the constant strain rate test at $\dot{\varepsilon}_3 = 6.7 \times 10^{-2}$ s^{-1} the serrated yielding effect is manifested as a discontinuity in the load (F)-elongation (Δl) curves (Fig. 1, curve 3) in the form of strain bursts, already observed by Hamersky,[6] indicating that the stress increase preceding the strain burst is of a plastic nature. The presence of plateaus on the strain-stress curves indicates type D serrations, first observed by Wijler *et al.*,[7] due to band propagation similar to Lüders band with no work hardening or strain gradient ahead of the moving band front,[45] because, outside the band, the dislocations are anchored by Mg solute atmospheres. A close inspection of the flow curve in Fig. 2 reveals that the plateaus are not completely flat. This indicates an interplay between strain hardening, due to an increase in the dislocation

density with deformation, and band propagation coupled with serrated yielding. If the plateaus are not completely flat, depending on which effect dominates the deformation process, the slopes of the plateaus can be positive or negative. A positive slope is caused by the dominant effect of homogeneous deformation and related strain hardening. A negative slope can be rationalized in terms of an increase in the vacancy concentration, which facilitates Mg diffusion and inhomogeneous deformation inside the propagating band.

According to Wijler *et al.*,[7] an increase of stress between two plateaus, $\Delta\sigma_b$ (see Fig. 2), can be correlated to the plateau length by the work hardening coefficient for homogeneous deformation, n, and the length of the plateau, $\Delta\varepsilon_b$:

$$\Delta\sigma_b = n \times \Delta\varepsilon_b \tag{8}$$

It is interesting to note that the strain hardening coefficient value, n (for curve 3 in Fig. 1, $n \approx 0.35$), is not the same as the average value of the strain hardening coefficients of the stress-strain segments on the same curve (Fig. 2), which is an average $n_{av} \approx 0.46$. This indicates significant pertubation of *PL* bands along the deformation curve, and represents only localized deformation behavior, which may or may not have a macroscopic effect in the form of mixed type serrations. This may be the explanation for the significant variations in the activation energy data for serrated yielding of Al-Mg type alloys present in the literature.[19,20,29-34,43] In addition, the influence of specimen geometry on the *PL* bands formation also has to be taken into account.[47]

SUMMARY

Serrated yielding in the investigated Al-5.6% Mg alloy was observed to be present from the onset of plastic deformation at room temperature, indicating that the critical temperature is below the test temperature and the point defect concentration is sufficiently high to allow the *PL* effect to appear with zero critical strain. Our calculations confirmed the experimental observations that, under the investigated conditions, serrated yielding would commence with zero critical strain. Regardless of the diffusion mechanism (volume or pipe), it is to be expected that, at room temperature, and the used strain rates, the Mg solute atoms in the Al matrix are so mobile that the majority of dislocations move only in conjunction with a Mg atmosphere.

The energy of activation for serrated yielding, E_{aSY}, is closely related and may be compared to the binding energy between the Mg solute and the dislocations, E_{bMgD}. However, is should be noted that the data available in the literature for the binding energy between vacancies and Mg solute, E_{bMgV}, the energy of activation for Mg solute diffusion in Al, E_{mMg}, as well as the energy of activation for serrated yielding of Al-Mg type alloys, E_{aSY}, vary considerably. Consequently, it is to be expected that considerations, regarding serrated yieldin in these alloys, based on the Mg solute/vacancy binding energy, and/or the binding energy between the Mg solute and the dislocations, may not be sufficiently conclusive and further work is needed in order to clarify these ambiguities.

With increasing strain rate, the appearance of serrated yielding changed from *C* to *A* or *A+B*, and finally to *D* type as a series of successive stress plateaus which are not completely flat (the stresses are not completely constant). The average strain hardening exponent of the idnividual plateaus differ from the *n* value estimated from the general flow curve. This effect is assumed to be the result of interplay between strain hardening and propagation of the proposed non-strain hardening deformation band in the materials with serrated yielding.

ИЗВОД

ДИСКОНТИНУИРАНО ПОПУШТАЊЕ КОД ЛЕГУРЕ AlMg6.5

ЕНДРЕ РОМХАЊИ, МИЉАНА ПОПОВИЋ и ВЕЛИМИР РАДМИЛОВИЋ

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Затезна испитивања су вршена на лиму од легуре AlMg6.5, након што је материјал хладно деформисан ваљањем 70% и жарен на 320 °C/3h. Брзине деформације при затезању су биле $6{,}7\times10^{-4}$ s^{-1}, $6{,}7\times10^{-3}$ s^{-1} и $6{,}7\times10^{-2}$ s^{-1}. Добијени резултати показују да се на собној температури, за све три брзине деформације, запажа појава Portevin-Le Chatelier-ов ефекта од самог почетка деформације. Другим речима, у свим случајевима критична деформација је једнака нули. Целовита анализа публикованих активационих енергија, релевантних за појаву дисконтинуираног попуштања, показују значајне разлике у вредностима. И поред тога, примењени модел израчунавања дифузионих коефицијената за дате услове експеримента, потврђује појаву дисконтинуираног попуштања при нултој критичној деформацији. Добијени резултати и примењени прорачун показују да су код легуре AlMg6.5, атоми магнезијума довољно покретни на собној температури, и да се већина дислокација креће заједно са атмосферама магенезијумових атома. Након највеће брзине деформације, запажена је појава "*D*" дисконтинуитета, као серије узастопних напонских платоа, указујући на сложени механизам деформације путем интеракције деформационог ојачавања и појаве деформационих трака које не ојачавају.

(Примљено 5. маја 1998)

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