INFLUENCE OF TEMPERATURE AND STRAIN RATE OF TENSILE TEST ON SERRATION IN 2017 ALUMINUM ALLOYS

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ABSTRACT Effects of strain rates and temperatures of the tensile test on serrated yielding were investigated in a commercial 2017 Al-Cu alloy solution treated at 773K and either quenched into water or cooled in the furnace. In water quenched specimens, serration was observed at temperatures between 243K and 333K. In the specimens cooled in the furnace, serration occurred at temperatures between 273K and 403K.

At low test temperatures, the critical strain, ε_c necessary to initiate serrated yielding increased remarkably with increasing strain rate, $\dot{\varepsilon}$. At high test temperatures, however, ε_c decreased significantly with increasing $\dot{\varepsilon}$. In both cases the relationship between ε_c and $\dot{\varepsilon}$ could be expressed as $\ln \varepsilon_c = a + b \ln \dot{\varepsilon}$, giving positive slope b at low test temperatures, and negative slope b at high test temperatures.

The activation energy calculated from these results are consistent with Cottrell's theory only in the low temperature test.

Keywords :*serrated yielding, Al-Cu alloy, activation energy, Cotterell's theory.*

1. INTRODUCTION

In Al-Mg alloys, serrated yielding has already been studied by many investigations [1 to 5]. However, although serrated yielding occurs also in Al-Cu alloys, little investigation has been made in these cases[6 and 7]. It is therefore tried to clarify serrated yielding in Al-Cu alloys systematically in detail in this investigation.

The results obtained were compared with those already reported in the case of Al-Mg alloys and differences in mechanisms between two alloys were discussed. In Al-Mg alloys, it has been reported that two types of serration, types "A" and "B" appear depending upon the test temperature[1]. Although the former can be satisfactorily explained with Cottrell's model[8], the latter has not found still an appropriate explanation. It was therefore tried to confirm whether similar phenomena can be observed also in Al-Cu alloys.

2. EXPERIMENTALS

The starting materials were commercial 2017 Al-Cu alloys. They were delivered as 1mm thick cold rolled sheets. Their chemical compositions are Cu:3.5%-4.5% Si<0.8% Zn<0.25% Mg:0.2-

0.8% Mn:0.4-1.0%. From these, tensile test specimens were machined according to standard JIS No.5 specification. They were annealed at 773K for 40min and either quenched into water (WQ) or cooled in furnace (FC).

Tensile tests were carried out on an Instron type tensile testing machine. Tensile tests were conducted at all test temperatures in an isothermal bath held at a constant temperature.

2.1 Influence of test temperature

The tensile test specimens were strained at constant strain rate of either 3.3×10^{-3} or $3.3\times10^{-4}sec^{-1}$. The test temperature was varied between 243K and 403K.

2.2 Influence of strain rate

Influence of strain rate was investigated at two test temperatures, i.e, at low and high test temperatures; In the case of water quenched specimens, tensile tests were performed at 263K and 313K, while, in the case of furnace cooled specimens, tensile tests were performed at 273K and 373K. Strain rates were changed between 3.3×10^{-5} and $3.3\times10^{-2}sec^{-1}$. From the recorder charts, strains, ε_c, necessary to initiate serrated yielding was determined.

3. RESULTS

3.1 Influence of test temperature

The strain to initiate serration, ε_c , is plotted in Figs.1 and.2 against the reciprocal of absolute test temperature.

In water quenched specimens, serration appeared at temperatures between 243K and 333K. In furnace cooled specimens, serration occurred at temperatures between 273K and 403K. In both cases, plots of ln ε_c versus 1/T yielded V-shaped curves. Type "A" serration occurred in the low test temperature range, whereas type "B" serration was observed in the high test temperature range. These results show that ln ε_c varies linearly with 1/T. In the case of the "A" type serration which was observed at low test temperature range, ε_c observed at high strain rate was grater than that observed at low strain rate. But in the case of the "B" type serration which was observed in the high test temperature range the results were reversed.

Thus increasing strain rate seems to be equivalent to elevate the test temperature.

3. 2 Influence of strain rate

Critical strain ε_c necessary to initiate serration is plotted in Figs.3 and 4 against strain rate $\dot{\varepsilon}$ in double logarithmic forms. It is evident that both in specimens quenched into water and in specimens cooled in the furnace, ln ε_c is proportional to $\dot{\varepsilon}$. But the slope of these plots is positive at low test temperatures, while it is negative at high test temperatures.

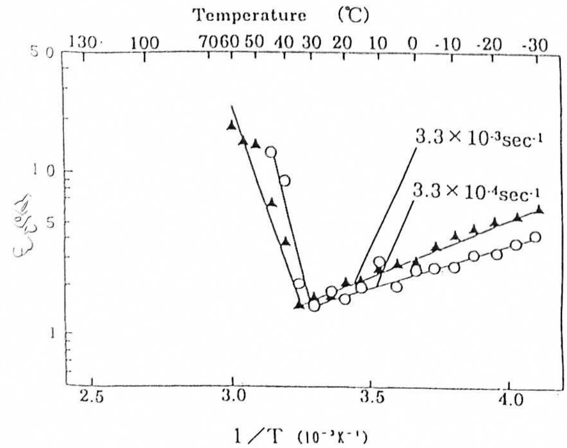

Fig.1 Relation between critical strain ε_c and test temperature, T. (Water quenched specimens)

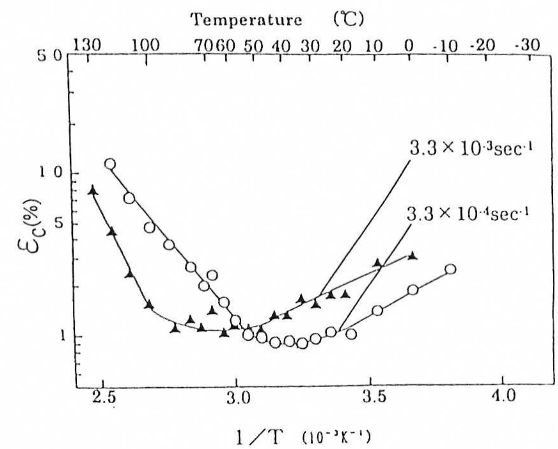

Fig.2 Relation between critical strain ε_c and test temperature, T. (Specimens cooled in the furnace)

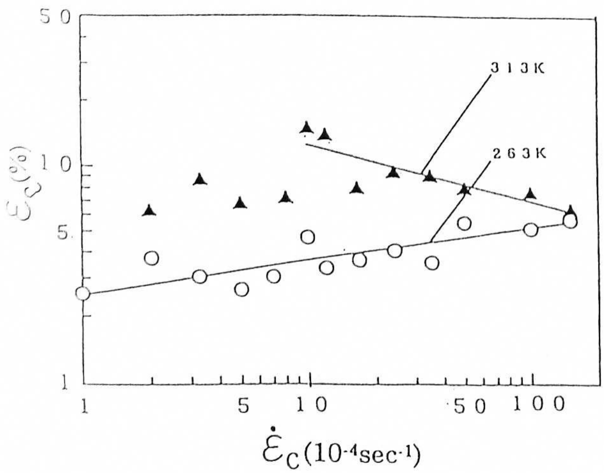

Fig.3 Relation between critical strain ε_c and strain rate $\dot{\varepsilon}$. Specimens quenched into water, tested at 263K and 313K.

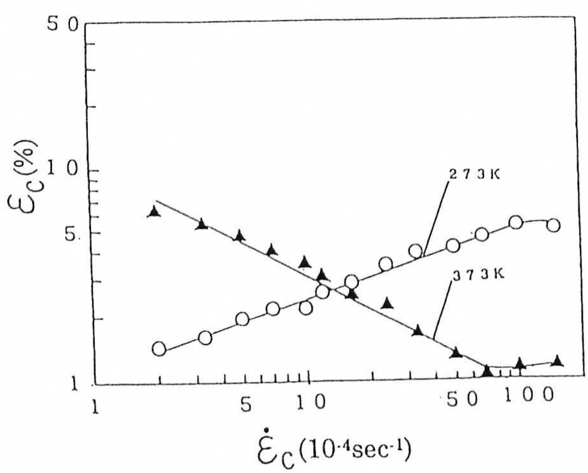

Fig.4 Relation between critical strain ε_c and strain rate $\dot{\varepsilon}$. Specimens cooled in the furnace, tested at 273K and 373K.

3.3 Activation energy

It is known that the dynamic solute-dislocation interaction or dynamic strain aging is the origin of serrated yielding, i.e, so called Portevin-Le Chatelier effect[9]. Impurity or solute atoms form an atmosphere around the moving dislocation, thereby making them immobile. The applied stress must therefore be increased, until it is sufficiently high to cause breakaway of the pinned dislocation. However, if the drift velocity of solute atoms is enough high, dislocations can be trapped by the solute atmosphere again. This process can be repeated many times before the final fracture of the samples occurs.

Cottrell[8] predicted that the serration can be initiated if the drift velocity of the solute atoms whose diffusion is assisted by vacancies, is equal to dislocation velocity. In that case, the diffusion coefficient D can be expressed by equation (1).

$$D=a^2 \nu ZC_v\exp(-E_m/kT) \qquad (1)$$

where a, ν, Z, C_v, E_m, k, and T are the interatomic distance, the Debye frequency, the coordination number of lattice, the vacancy concentration, the effective vacancy migration energy in the alloy, Boltzman's constant, and absolute temperature, respectively.

Concentration of vacancies C_v generated during plastic deformation can be related to the plastic strain ε by the empirical relation Eq (2).

$$C_v=K \varepsilon^m \qquad (2)$$

where K and m are material constants.

The solute atoms atmosphere will be formed at a critical dislocation velocity V_c given by Eq.(3).

$$V_c=4D/L \qquad (3)$$

where L is the effective radius of solute atom atmosphere formed about a dislocation. Strain rate $\dot{\varepsilon}$ is related with average dislocation velocity V_c in the form given by Eq.(4).

$$\dot{\varepsilon}=\rho bV \qquad (4)$$

where ρ and b are density of mobile dislocations and magnitude of Burgers vector.

The critical strain rate $\dot{\varepsilon}_c$ can be derived from the condition $V=V_c$,

$$\dot{\varepsilon}_c=b\rho V_c=(4b\rho/L) \qquad (5)$$

Substitution of Eqs.(2) and (4) into Eq.(1) gives the following relationship.

$$\dot{\varepsilon}_c=(4b\rho/L)\cdot a^2 \nu ZK \varepsilon_c^m \cdot \exp(-E_m/kT) \qquad (6)$$

Densities of mobile dislocations, ρ can be given by the following equation

$$\rho=N\cdot\varepsilon^{\beta} \qquad (7)$$

where N and β are constants. By putting $(4a^2 \nu ZNbK)/L \equiv \theta$,we finally obtain

$$\dot{\varepsilon}_c = \theta \cdot \varepsilon_c^{(m+\beta)} \cdot \exp(-E_m/kT) \quad (8)$$

4. Discussions

Figures.1 and.2 show that that for both type "A" serration and type "B" serration, $\ln \dot{\varepsilon}_c$ is proportional to the reciprocal of the absolute temperature. Also Figs.3 and.4 show that $\ln \varepsilon_c$ is linearly related to $\dot{\varepsilon}_c$.

In these discussions, it has been assumed that dislocation velocity is equal to the drift velocity of solute atoms, Eq.(8). By taking logarithms of both sides of Eq.(8), we can obtain:

$$\ln \dot{\varepsilon}_c = \{(m+\beta)\ln \varepsilon_c\} - (-E_m/kT) + \ln \theta \quad (9)$$

For tensile tests performed at constant temperatures, 1/T is constant: Eq.(9) can be expressed therefore in the following from.

$$\ln \varepsilon_c = (m+\beta)\ln \varepsilon_c + A \quad (10)$$

where A represents a constant. From the slope of the $\ln \dot{\varepsilon}_c$ versus $\ln \varepsilon_c$ plot the value of $(m+\beta)$ can be obtained. For tensile tests performed at a constant strain rate, $\dot{\varepsilon}_c$ in Eq.(9) is constant. Then we obtain.

$$\ln \varepsilon_c = \{1/(m+\beta)\} \cdot (E_m/kT) + B \quad (11)$$

Activation energy E_m obtained from slope of the $\ln \varepsilon_c$ versus 1/T plot is given in Table 1. As to the "A" type serration, activation energy was about 0.6eV, except the value of 0.89eV. This activation energy seems to be nearly equal to the migration energy of vacancy in pure aluminum. As to the "B" type serration, on the other hand, the activation energy is largely different from the migration energy of vacancy in pure aluminum. Thus, these results shows that "B" type serration observed at high test temperatures cannot be explained in terms of Cottrell's model, as it has been observed also in Al-Mg alloys.

Table 1. Activation energy [eV]

	low test temperature range		high test temperature range	
	$3.3\times10^{-4}sec^{-1}$	$3.3\times10^{-3}sec^{-1}$	$3.3\times10^{-4}sec^{-1}$	$3.3\times10^{-3}sec^{-1}$
WQ	0.67	0.89	5.09	3.17
FC	0.60	0.50	0.74	0.13

In the case of the "B" type serration, dynamic recovery seems to play also an important role. In the high temperature test, the plastic strain for the initiation of the "B" type serration is larger at higher test temperatures and at slower strain rate. Precipitation of Cu atoms in the form of the Guinier-Preston zone many affects densities and mobility of dislocations. Decreases in the dislocation velocity seems to suppresses the initiation of the "B" type serration.

5.Conclusions

Influences of test temperatures and strain rates on serrated yielding have been studied in commercial 2017 aluminum alloys and following conclusions were obtained.

At low test temperatures, critical strain ε_c to initiate serration decreased with increasing test temperature. On the other hands, at high test temperatures ε_c increased with increasing test temperature. As to the relation between the strain rate $\dot{\varepsilon}$ and ε_c, ln $\dot{\varepsilon}$ versus ln ε_c plot showed positive slope at low test temperatures. But at high test temperatures, these plots yielded negative slopes, as it has been observed in Al-Mg alloys.

Activation energy showed that "A" type serration which is observed at the low test temperature range, can be explained in terms of Cottrell's model, but "B" type serration which is obsrved at high test temperature range, cannot be explained with this mechanism.

7. References

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