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Mechanical Equation of State in an Al-Li Alloy

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Abstract**—**Existence of plastic equation of state has been investigated by performing a series of load relaxation tests at various temperatures using an Al-Li alloy. A plastic equation of state is first developed from a simple kinetics consideration for a mechanical activation process of a leading dislocation piled up against grain boundaries. A series of load relaxation test has been conducted at temperatures ranging from 200 to 530°C to obtain the stress-strain rate curves. A plastic equation of state has been derived from a simple consideration of dislocation kinetics and confirmed by experimental results.

Keywords**—**Plastic equation of state, Dislocation kinetics, Load relaxation test, Al-Li alloy, Microstructure.

I. Introduction

EXISTENCE of plastic equation of state has been investigated by performing a series of load relaxation tests at various temperatures on an Al-Li alloy. The purpose of this work is to demonstrate that commercial Al alloy also exhibit the plastic equation of state [1], [2] and subsequently demonstrated for the various metals and alloys having a cubic structure such as pure Al [3], Pb [4] and type 300 stainless steels [5]. The concept of mechanical equation of state was first introduced by Hart [1] from the observation of a certain scaling relation of stress-strain rate curves obtained at various plastic strain levels for a particular material [2]. It was suggested that a plastic equation of state could be formulated in terms of stress, plastic strain rate, temperature and a hardness parameter. A new comprehensive approach has been proposed utilizing a simple dislocation dynamics by Chang and Aifantis [6] to provide a general framework for an internal variable theory of inelastic deformation pertaining to crystalline materials. They have used the concept of an imaginary internal observer within a crystal, through whom the actual internal deformation processes can be monitored. In this way, the role of defect species, say dislocations, that are responsible for many macroscopic processes such as creep deformation, shear localization, and crack propagation can effectively be described in contrast to the usual phenomenological approach by an external observer.

In this study, a plastic equation of state is first introduced from a simple kinetics consideration for a mechanical activation process of a leading dislocation piled up against grain boundaries. A series of load relaxation test has been conducted at temperatures ranging from 200°C to 530°C to obtain the stress-strain rate curves. The results are then examined in connection with the proposed plastic equation of state.

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II. Plastic Equation of State

An elementary material volume V bounded by a surface S is considered here, within and across which dislocations are allowed to move in response to an applied stress. When the material boundary acts a strong barrier partially blocking the passage of dislocations, some of them would remain inside a material volume giving rise to an internal strain ($\dot{\alpha}$), while the rest will pass through to produce an externally observable plastic strain rate ($\dot{\alpha}$). The total inelastic strain rate ($\dot{\varepsilon}$) can be, therefore, expressed as the sum of the internal strain rate ($\mathring{a}$) and non-recoverable plastic strain rate ($\dot{\alpha}$) in the temperature range, where other deformation mechanisms such as grain boundary sliding do not operate significantly. The applied stress may also be decomposed into an internal stress σ^I and a friction stress σ^F [7]. The former is responsible for driving dislocations against the long-range interaction forces due to piled-up dislocations and the latter is used for dislocations to overcome the frictional resistance of lattice [7]. Thus we have the following stress relation and kinematic relation among the deformation rate variables $\mathring{a}$,$\dot{\alpha}$ and $\dot{\varepsilon}$ in a tensorial form,

$$\sigma = \sigma^I + \sigma^F \tag{1}$$

$$\dot{\varepsilon} = \mathring{a} + \dot{\alpha} \tag{2}$$

The time rate $\mathring{a}$ denotes the co-rotational rate of internal strain tensor spinning with an internal observer, similar but somewhat different to the material time rate used by Hart [8]. A rheological model can be constructed from the above two relations by identifying the proper correspondence of stress and strain rate variables used in (1) and (2).

Since the test temperatures of this study corresponds to a high homologous temperature of above $0.5T_M$ for Al alloy, σ^F is in general very small compared to σ^I and $\mathring{a}$ can be neglected if the relaxation test is performed uniaxially at a steady state, It is therefore sufficient to describe the constitutive relation for $\dot{\alpha}$ element at temperatures employed in this study. The constitutive relation between σ^I and $\dot{\alpha}$ under a uniaxial condition can be obtained from a simple consideration of dislocation kinetics as follows.

The leading dislocation in a pile-up is thought to overcome strong barriers such as grain boundary by a mechanical activation process. The dislocation flux j^B across the barriers can, therefore, be given in a form,

$$j^B = j^B (f^I / f^* ; T; \text{Barrier structure}) \tag{3}$$

with f^I denoting a driving force and f^* a resisting force. With the aid of physical conditions that must be satisfied by a mechanical activation process given by;

$$j^B = 0 \text{ when } f^I = 0,$$

$$j^B \to \infty \text{ when } f^I \to f^* \quad (4)$$

The following kinetics relation can be prescribed;

$$\ln (f^*/f^I) = (j^*/j^B)^p, \quad (5)$$

with p denoting a material constant and j^* a reference flux at $(f^*/f^I) = e$. When the above kinetics relation is expressed in terms of stress and strain rate variable, a "plastic equation of state" can be obtained as

$$\ln (\sigma^*/\sigma^I) = (\dot{\alpha}^*/\dot{\alpha})^p, \quad (6)$$

similar to that proposed phenomenologically by Hart [8]. The temperature dependence of this mechanical activation process is then prescribed through the $\dot{\alpha}^*$ variable as

$$\dot{\alpha}^* = \omega^I(\sigma^*/\mu)^{n^I}\exp(-Q^I/RT), \quad (7)$$

with ω^I denoting a jump frequency, Q^I an activation energy, and μ an internal modulus, R the gas constant, and n^I a material constant [9].

III. Experimental Procedures

TABLE I
CHEMICAL COMPOSITIONS OF THE AL-LI ALLOY USED IN THIS STUDY

Cu	Fe	Mg	Si	Zr	Zn	Li	Al
1.24	0.05	0.9	0.02	0.11	0.04	2.5	Bal.

The 8090 A1-Li alloy used in this study was supplied in the form of a plate with the thickness of 44.5 mm, solution treated, 6% stretched and naturally aged condition (T3771). This material, known as LITHAL together with 8091 Al-Li alloy in generic name, had the chemical composition as given in Table I. A proper thermomechanical treatment (TMT) is required before expecting the superplasticity for this alloy because the as-received material has elongated and pancake-shape microstructure. The basic concept of thermomechanical treatment (for grain refinement of precipitation hardening aluminum alloy) proposed by Wert et al. [9] was employed with some modifications in this study.

Load relaxation tests were then carried out at the various temperatures between 200°C and 530°C by using a computer controlled electro-mechanical testing machine (Instron 1361 model) attached with a furnace capable of maintaining the temperature fluctuation within ±1°C. The variation of load with time was monitored through a DVM and stored in a personal computer for subsequent data analysis to determine the flow stress σ and inelastic strain rate $\dot{\varepsilon}$ following the usual procedure described in the literature [10].

IV. Results and Discussion

The relaxation test results obtained from the two successive runs at 200°C are given in Fig. 1. The curve A represents an initial relaxation result, while B was obtained from the second and third run after reloading the same specimen above the peak load of previous run, respectively. The total prior plastic strains were 1.2 per cent for A and2.6 per cent for B. The initial loading strain rate was 1.0×10^{-2}/sec. At this temperature log σ vs. log $\dot{\varepsilon}$ curves effectively represent the relation between σ^I and $\dot{\alpha}$ as stated previously. The constitutive parameters in (6) were then determined by using a non-linear regression method and the results are given in Table II.

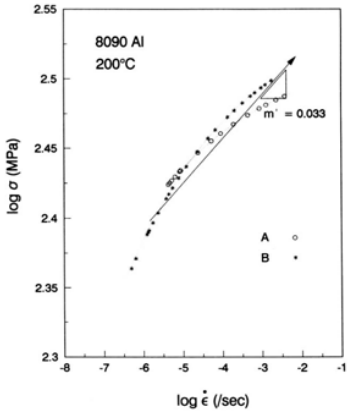

Fig. 1 The scaling relation of the two curves representing the plastic behavior

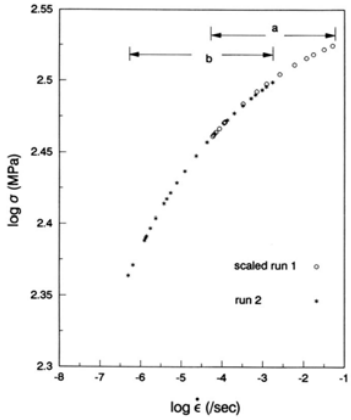

Fig. 2 The master curve for 8090 Al-Li alloy generated from the two curves in Fig. 1. The ranges denoted as a and b show the extent of each curve

The solid lines in Fig. 1 are, in fact, the predicted curves from (6) using the constitutive parameters listed in Table II. It is noted that the value of exponent p is obtained as 1. It is noted that the value of exponent p is obtained as 0.15, the same as previously results for cubic crystals [11]. This is believed to come from the fact that the parameter P characterizes the dislocation permeability through strong barriers, say grain boundaries, and thus depends strongly on the barrier properties

including its geometry and structure [6]. Some previous observations [12] show that grain boundary structure of a BCT tin is quite different from those of cubic metals and the value of this parameter was obtained as 0.1.

It is interesting to note from Fig. 3 that microstructure after two successive load relaxation tests seemed very similar to that before the tests.

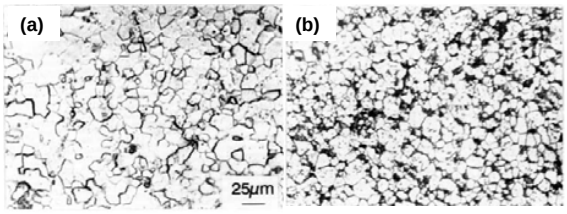

Fig. 3 Optical microstructure taken from the specimen (a) before and (b) after load relaxation test at 200°C as in Fig. 1

TABLE II
THE CONSTITUTIVE PARAMETERS DETERMINED FROM THE LOAD RELAXATION TESTS AT 200°C.

Test	log σ*(MPa)	log $\dot{\alpha}$*	p
A	2.523	-9.624	0.15
B	2.554	-8.674	0.15

The arrow marked in Fig. 1 represents the scaling relation that can be derived by differentiating (6) and (7) as

$$\left.\frac{\partial \ln \sigma^*}{\partial \ln \dot{\alpha}^*}\right|_T = \left.\frac{\partial \ln \sigma^*}{\partial \ln \dot{\alpha}^*}\right|_{v^I} = \frac{1}{n^I} = m^I, \quad (8)$$

$$\left.\frac{\partial \ln \sigma^I}{\partial \ln \dot{\alpha}}\right|_{T,\sigma^*} = v^I \quad (9)$$

In fact, this scaling relation provides a means to make the three curves coincide by a translation along a straight line with the slope, m^I. The master curve obtained in this way is shown in Fig. 2 implying that (6) is indeed a mechanical equation of state.

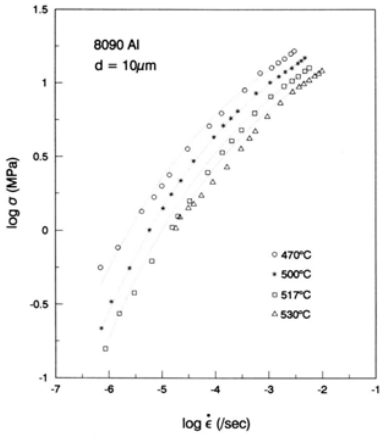

Fig. 4 The effect of temperature on the flow behavior of 8090 Al-Li alloy

The flow curves obtained from relaxation test results at four different temperatures are shown in Fig. 4. The values of constitutive parameters calculated from a non-linear regression method are presented in Table III. The solid lines are again the predicted curves from (6) with the values of constitutive parameters listed in Table III. Since the flow curves given in Fig. 4 show a similar level of internal strength variable σ^* within the test temperature range as can be seen from Table III, the activation energy Q^I prescribed in (7) can now be determined from the obtained by differentiating (7) at a constant σ^*. The activation energy was obtained as 130kJ/mol, which is somewhat larger than the activation energy for the self-diffusion of pure Al [13].

TABLE III
THE CONSTITUTIVE PARAMETERS DETERMINED FROM THE LOAD RELAXATION TESTS AT TEMPERATURES FROM 470°C TO 530°C.

Temp.(°C)	log σ*(MPa)	log $\dot{\alpha}$*	p
470	1.87	-1.47	0.15
500	1.83	-1.07	0.15
517	1.77	-0.92	0.15
530	2.5541.74	-0.79	0.15

V. CONCLUSIONS

By conducting a series of load relaxation tests on 8090 Al-Li alloy in connection with the proposed plastic equation of state, the following important results were obtained.

1. A plastic equation of state has been derived from a simple consideration of dislocation kinetics as

$$\ln (\sigma^* / \sigma^I) = (\dot{\alpha}^* / \dot{\alpha})^p,$$

2. The plastic flow curves of log σ^I vs. log $\dot{\alpha}$ can precisely be described by the proposed plastic equation of state at temperatures ranging from 200°C to 530°Cwith the value of permeability parameter p = 0.15.
3. The activation energy Q^I for the mechanical activation process of a leading dislocation at grain boundaries of white tin has been obtained as 130kJ/mol, which is somewhat smaller than that of self-diffusion inpure Al.

ACKNOWLEDGMENT

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