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EFFECT OF VACANCY SINKS AND SOURCES ON SERRATED YIELDING DUE TO SOLUTE LOCKING

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It is generally accepted that a realistic model for the Portevin-Le Chatelier effect due to excess vacancy stimulated solute diffusion and dislocation locking must consider vacancy annihilation as well as vacancy creation during straining (1). Modified versions (2,3) of Cottrell model (4) of repeated yielding take into account the thermal vacancy contribution as well as the strain-produced vacancies but fail to consider annealing of vacancies to appropriate sinks. Recent experimental data on Al-Mg alloy by MacEwen and Ramaswami (3) reveal important deviations from Cottrell model which they qualitatively accounted for by vacancy annihilation. The present note is an attempt to quantitatively explain these deviations by taking into account the disappearance of vacancies at sinks.

The Cottrell model has been most used and abused in attempts to explain the experimental findings of the delay of plastic strain (ε_o) and critical strain-rate ($\dot{\varepsilon}_c$) for the appearance of serrations and their temperature dependencies. According to Cottrell (4) the critical strain-

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rate at which serrations start appearing is given to be

$$\dot{\varepsilon}_c = K \rho_m C_v e^{-E_m/RT} , \qquad [1]$$

where K is a constant, ρ_m the mobile dislocation density, C_v the vacancy concentration and E_m the activation energy for vacancy migration. Except for two cases (2,3) C_v in the above equation was taken to be the total concentration of vacancies produced during straining whereas in strict sense C_v is the <u>net</u> total vacancy concentration and

$$C_v = C_v^T + C_v^\varepsilon - C_v^- , \qquad [2]$$

Where C_v^T is the thermal equilibrium vacancy concentration at the test temperature T, C_v^ε the total excess vacancy concentration produced by straining at strain ε and C_v^- the concentration of annihilated vacancies at temperature T, and $C_v^- \leq C_v^\varepsilon$.

Point defects are generated by the non-conservative motion of jogs on screw dislocations (5) and the defects thus produced are vacancies or interstitials depending upon the nature of the jogs. Assuming that vacancies are created at vacancy producing jogs on gliding screws then the rate of production of these excess vacancies is given by

$$\dot{C}_v^\varepsilon = \frac{\rho_s v_s p}{N l_j b} ,$$

where ρ_s is the density of mobile screw dislocations, v_s the velocity of these screw dislocations, p the height of jogs in Burgers vectors, l_j the jog separation, b the Burgers vector and N the number of lattice sites per unit volume. Using $\dot{\gamma} = \frac{3}{4} \dot{\varepsilon} = \rho_s b v_s$ we find (6)

$$\dot{C}_v^\varepsilon = \frac{3}{4N} \left(\frac{p}{l_j b^2} \right) \dot{\varepsilon} , \qquad [3]$$

where $\dot{\varepsilon}$ is the imposed tensile strain-rate. Thus if there is no recovery, i.e. once formed a vacancy does not anneal out to a sink, we find from Eq. 3 that

$$C_v^{\varepsilon} = \frac{3}{4N}\left(\frac{\rho}{l_j\, b^2}\right)\varepsilon \;, \qquad [4]$$

In reality however such excess vacancies disappear at sinks at a rate proportional to the number produced so that

$$\dot{C}_v^{-} = \left(\frac{4D_v}{L_s^2}\right) C_v^{\varepsilon} \;, \qquad [5]$$

where D_v is the vacancy diffusivity and L_s the vacancy source-sink distance. In a very fine grained material this L_s may be taken as the grain size, but in general dislocations are the primary vacancy sinks in which case

$$L_s^2 \simeq \frac{1}{\rho} \;, \qquad [6]$$

so that

$$\dot{C}_v^{-} = 4D_v\, \rho\, K_1\, \varepsilon \;, \qquad [7]$$

with $K_1 = \frac{3}{4N}\frac{\rho}{l_j\, b^2}$. It has been well established (7) that the density of dislocations, ρ, varies with the square of the flow stress and $\sigma = \alpha\, Gb\sqrt{\rho}$ with $\alpha \sim 1$. On the other hand $\sigma \equiv \sigma\{\varepsilon\}$ dependent on conditions and in general $\sigma = A\sqrt{\varepsilon}$ so that $\rho = \rho_o \varepsilon$ while in single crystals for Stage II deformation $\sigma = B\,(\varepsilon - \varepsilon_o)$ so that $\rho \sim \varepsilon^2$. Substituting the explicit strain-dependence of the dislocation density and integrating we obtain

$$C_v^{-} = \frac{4\, D_v\, K_1}{3\, \dot{\varepsilon}}\; \rho\, \varepsilon^2 \qquad [8]$$

A similar but more complicated expression may be obtained for single crystals. Thus the net total vacancy concentration present at strain ε and temperature T is

$$C_v = C_v^T + C_v^\varepsilon - C_v^- = e^{-(E_f + TS_f)/RT} + \frac{3\ p}{4N1_j\ b^2}\left(\varepsilon - \frac{4\ D_v\rho}{3\ \dot{\varepsilon}}\ \varepsilon^2\right) \qquad [9]$$

where E_f and S_f are the activation energy and entropy respectively of vacancy formation.

As an illustration of the applicability of the above analysis recent data of MacEwen and Ramaswami (3) on serrated yielding in Al-Mg alloy will be considered. Fig. 1 is a reproduction of their data on the temperature dependence of ε_o. The datum point in this figure at $\varepsilon_o = 1.23 \times 10^{-3}$ falls above the straight line obtained from other points based on Cottrell equation. If the vacancy annihilation at the corresponding temperature (370°K) were negligible the datum point would have been at $\varepsilon_{new} = 5.0 \times 10^{-4}$. Because some vacancies annealed out it needed more than this ε_{new}. Hence $C_v^\varepsilon\{\varepsilon_o\} - C_v^-\{\varepsilon_o\} = C_v^\varepsilon\{\varepsilon_{new}\}$. Or, from Eqs. [4] and [9]

$$C_v^\varepsilon\{\varepsilon_{new}\} = K_1\ \varepsilon_{new} = K_1\left[\varepsilon_o - \frac{4\ D_v\rho}{3\dot{\varepsilon}}\ \varepsilon_o^2\right] \qquad . \qquad [10]$$

Inserting the experimental values of $\dot{\varepsilon}$, ε_o, D_v at 370°K and the extrapolated value of ε_{new}, a value of 2×10^9 cm^{-2} is obtained for the dislocation density. The value for ρ is of the same order (indeed 3 times) as expected from Ham-Jaffrey (8) equation,

$$\rho = 4.73 \times 10^{11}\ \varepsilon = 6 \times 10^8 \text{ for } \varepsilon = 1.23 \times 10^{-3} \quad .$$

Thus the present analysis yields excellent agreement with the experimental findgs. With the above modifications then the Cottrell equation for the critical (imposed) strain-rate may be rewritten to be

$$\dot{\varepsilon}_c = K'\ \varepsilon_o\ e^{-E_m/RT}\left\{e^{-E_f/RT} + \frac{3\ p\ \varepsilon_o}{4N1_j b^2}\left(1 - \frac{4\ D_v\rho}{3\dot{\varepsilon}_c}\ \varepsilon_o\right)\right\}$$

in lieu of the simplified equation used thus far.

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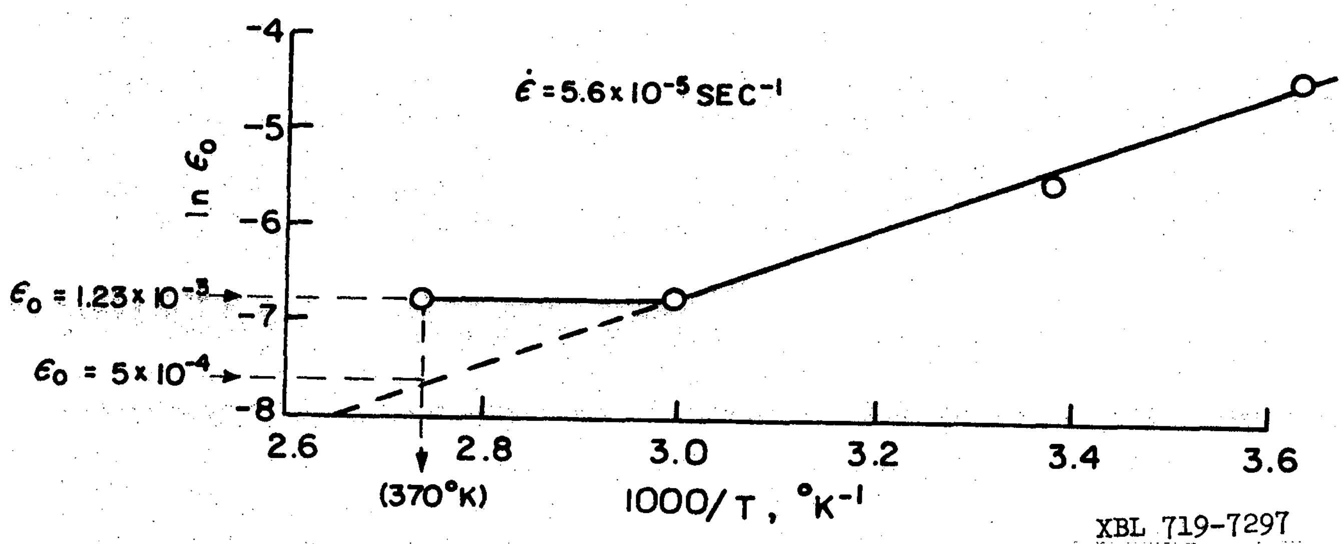

FIG. 1

Plot of ln ϵ_o vs 1/T. Data of MacEwen and Ramaswami. (3)

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