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Work hardening of mild steel within dynamic strain ageing temperatures

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An investigation has been performed on the plastic behaviour of a mild steel within the region of dynamic strain ageing. For this purpose tension tests have been performed on annealed XC18 steel within a range of temperatures, from 305–776 K, and a range of strain rates, from 1.0×10^{-4}–1.85×10^{-1} s^{-1}. An analysis of experimental results is presented using a model for plastic deformation based on dislocation multiplications.

1. Introduction

Dynamic strain ageing is common in BCC metals containing interstitial atoms. The presence of dynamic strain ageing is manifested in serrated stress–strain curves, increased work-hardening and low ductility. In addition, in certain conditions a negative rate-sensitivity as determined at constant strain rates may be observed.

It is well recognised that all effects associated with dynamic strain ageing are caused by interstitial atoms diffusing during plastic deformation into gliding dislocations or by stress-induced ordering of solute atoms [1]. Cottrell atmospheres (diffusion induced) or Snoek atmospheres (stress induced) develop so called drag during dislocation glide which contributes to some extent to an increased flow stress. These effects occur for mild steels in the approximate temperature range $350\text{ K} < T < 600\text{ K}$. Transmission electron microscopy (TEM) studies seem to indicate that an increased mean work hardening rate $\theta = d\sigma/d\varepsilon$ in the dynamic strain ageing region of T and $\dot{\varepsilon}$ is caused by an enhancement of dislocation multiplication rate due to locking of mobile dislocations rather than by the dragging of atmospheres [2]. This point of view is now commonly accepted [3] for the case of interstitial alloys.

In this communication a unified dislocation model previously developed to describe the stress–strain behaviour of polycrystalline BCC metals and alloys is applied to analyse dynamic strain ageing in a mild steel XC18 (French Standards), [4–7]. By evaluating three material constants introduced in this model, which describe strain hardening via experimentally determined stress–strain curves at various temperatures in the dynamic ageing range, it is possible to find a good physical interpretation of the process.

The analysis is based on the constitutive formalism outlined in several publications [4–8]. It is allowed in the formalism that not only the flow stress, but also the rate of microstructural evolution with strain depends on strain rate and temperature. Such a point of view is also represented by others [8, 9]. A unified theoretical concept is employed in which rate-sensitive strain hardening, temperature and rate-sensitivity are included in terms of kinetics of dislocation multiplication, glide and annihilation. The flow stress σ at constant structure is well approximated by:

$$\sigma(\dot{\varepsilon}, T)_{STR} = \sigma_\mu \{s_j [h(\dot{\varepsilon}, T)]\}_{STR} + \sigma^* \{s_j [h(\dot{\varepsilon}, T)], \dot{\varepsilon}, T\}_{STR} \quad (1)$$

Where σ_μ and σ^* are respectively the internal and effective stress components and $h(\dot{\varepsilon}, T)$ is the thermal-mechanical history. The dislocation kinetics are associated with σ^* through an Arrhenius equation for dislocation glides [4–7] whereas σ_μ is associated with microstructural evolution of strain hardening. The internal stress σ_μ must be also rate, time and temperature dependent via dynamic recovery and dynamic ageing processes. Both components of stress in Equation 1 are written for a current state characterized by s_j state variables [6, 7]. Since one-parameter models based on the evolution of mean dislocation density ρ have proved their usefulness in constitutive modelling, the simplest evolutionary relation will be used in this analysis, [10].

$$\frac{d\rho}{d\varepsilon} = M - K_a(\dot{\varepsilon}, T)(\rho - \rho_0) \quad (2)$$

Where M is the multiplication factor, $K_a(\dot{\varepsilon}, T)$ is the annihilation factor and ρ_0 is the initial dislocation density. The internal stress σ_μ is related to the mean dislocation density ρ by the standard relation:

$$\sigma_\mu = \alpha\mu(T)\, b\, \rho^{1/2} \quad (3)$$

Where α is a constant which characterizes the "strength" of obstacles to dislocation movement, μ is shear modulus at temperature T, b is the magnitude of Burgers vector. Integration of Equation 1 with initial

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conditions $\rho = \rho_0$ for $\varepsilon = 0$ yields the following result:

$$\rho = \rho_0 + \frac{M}{K_a(\dot{\varepsilon}, T)} \{1 - \exp[-K_a(\dot{\varepsilon}, T)\varepsilon]\} \quad (4)$$

Introduction of Equation 4 into Equation 3 provides the simplest constitutive relation with three material constants α, M_0 and K_a.

$$\sigma_\mu = \alpha\mu(T)\, b \left\{\rho_0 + \frac{M}{K_a(\dot{\varepsilon}, T)}[1 - \exp(-K_a\varepsilon)]\right\}^{1/2} \quad (5)$$

It is assumed in this simple model that the mean free path λ of dislocations is constant, i.e. $M = 1/b\lambda$. The effective stress component σ^* is intentionally ignored in this analysis because in the dynamic strain ageing domain its value is negligible. It is result of conclusion that the Cottrell drag does not contribute much to the flow stress [2, 11, 12].

2. Experimental procedure

The material used in the tensile testing was XC18 mild steel. This is a calm steel with 0.18%C, defined by the French Standard NF A.02-005, with the composition shown in Table I.

Cylindrical specimens of diameter 5 mm and active length 40 mm were cut from a hot rolled bar. After machining, all specimens (total number 140) were annealed at 937 K for 4 h. The mean grain size after heat treatment was 15 μm.

Tensile tests were performed at 11 temperatures within the following range $305\ \text{K} < T < 776\ \text{K}$ at ~ 50 K steps. The temperature was controlled within ± 5 K. At each temperature four strain rates $\dot{\varepsilon}$ were applied $\dot{\varepsilon}_1 = 1 \times 10^{-4}$; 2×10^{-3}; 2×10^{-2}; 1.85×10^{-1} s^{-1}. For each condition of experiment $(T, \dot{\varepsilon})$ three specimens were tested. A closed-loop, computer controlled, screw testing machine was used.

Test results in the form of force–displacement curves were recorded in digital form on the disketts and this data was used in the detailed analyses.

3. Results

The schematic $\sigma(\varepsilon)$ curve shown in Fig. 1 illustrates characteristic points determined from each test, where σ_e is the upper yield limit, σ_e' the yield limit at which strain hardening starts, σ_m is the flow stress at maximum force, ε_L is the Lüders strain and ε_m is the strain at maximum force.

In order to determine material constants α, M and K_a Equation 5 has been used. The constant $\alpha(T)$, which is related to dislocation/obstacle interaction, can be determined at $\varepsilon = 0$ and $\rho = \rho_0$ by substituting

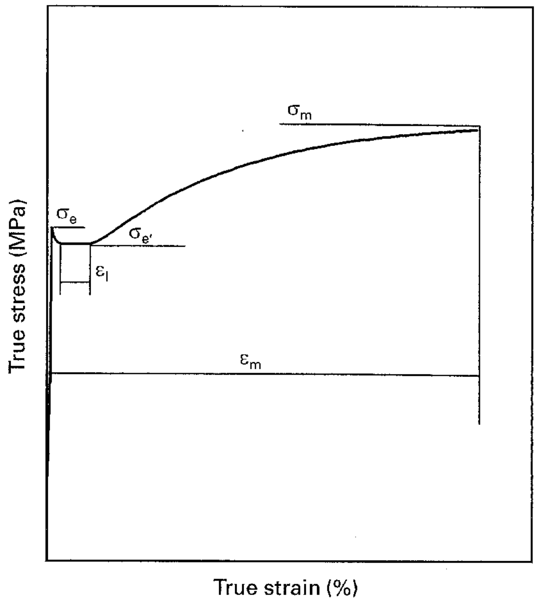

Figure 1 Characteristic points determined from each $\sigma(\varepsilon)$ curve, $T = \text{const.}$, $\dot{\varepsilon} = \text{const.}$

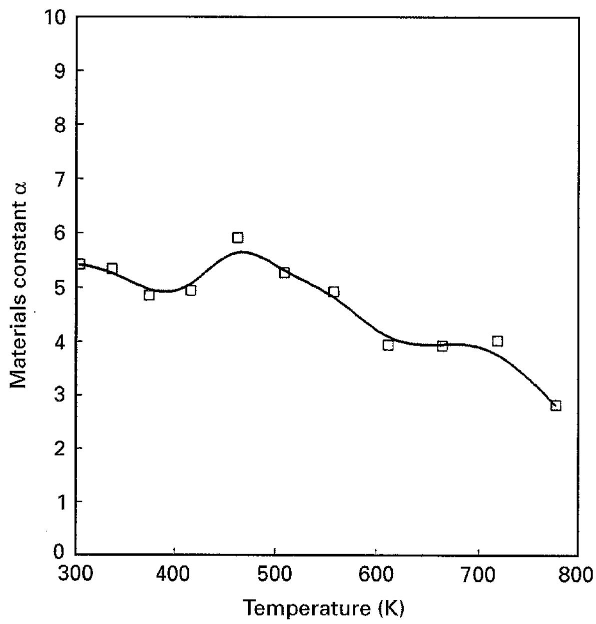

Figure 2 Changes of $\alpha(T)$ as a function of temperature. The strain rate used was 1.0×10^{-4}.

Equation 5 into Equation 6

$$\alpha(T) = \frac{\sigma_e(T)}{b\,\mu(T)\,\rho_0^{1/2}} \quad (6)$$

Changes of shear modulus $\mu(T)$ as a function of T were taken from experimental data, the initial dislocation density $\rho_0 = 6.2 \times 10^{-8}\ \text{cm}^{-2}$. Determined values of $\alpha(T)$ are shown in Fig. 2. A special extrapolation procedure which eliminates all rate effects

TABLE I Composition of the XC18 steel

	C	Mn	Si	S	P
standard composition %	0.15–0.22	0.5–0.8	< 0.35	< 0.035	< 0.04
actual composition %	0.17	0.58	0.21	0.032	0.024

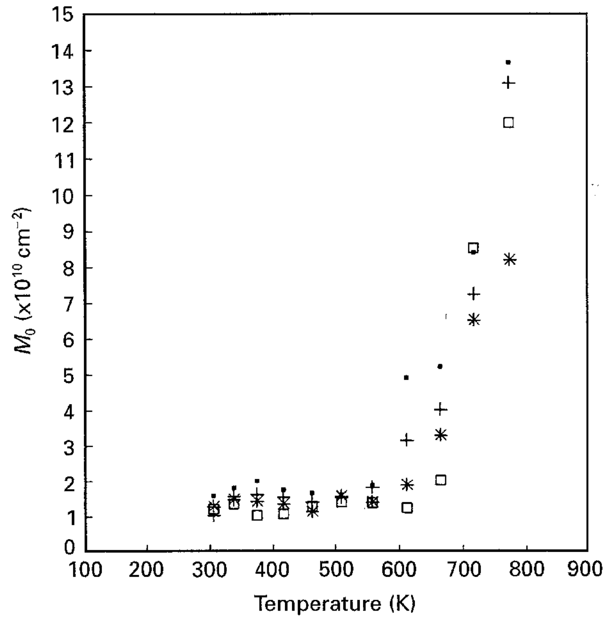

Figure 3 Determined values of M as a function of temperature T for four strain rates $\dot{\varepsilon}$ of (■) 1×10^{-4}, (+) 2×10^{-3}, (*) 2.0×10^{-2} and (□) 1.85×10^{-1}.

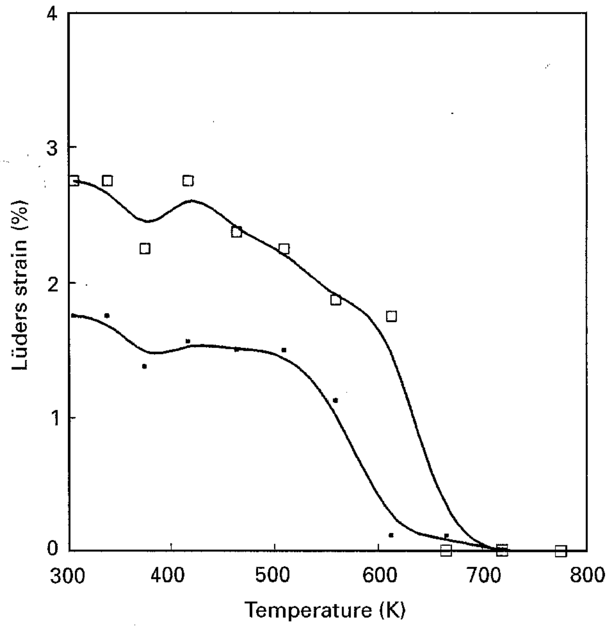

Figure 5 Values of Lüders strain as a function of temperature T for two extreme strain rates $\dot{\varepsilon}$ of (■) 1×10^{-4} and (□) 1.85×10^{-1}.

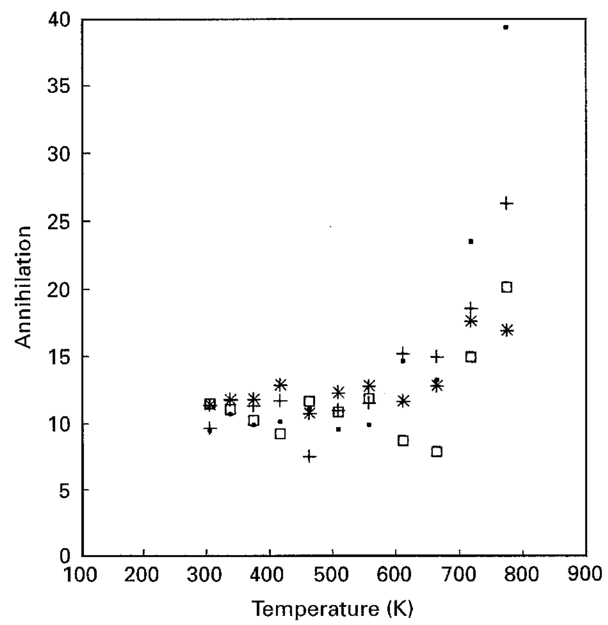

Figure 4 Determined values of K_a as a function of temperature T for four strain rates $\dot{\varepsilon}$ of (■) 1×10^{-4} (+), 2×10^{-3}, (*) 2.0×10^{-2} and (□) 1.85×10^{-1}.

has been applied in the determination of $\alpha(T)$, as a result each point in Fig. 2 is the mean of 12 tests.

In order to determine M and K_a as a function of T the entire curve $\sigma(\varepsilon)_{\dot{\varepsilon},T}$ had to be analysed and an optimization procedure must be applied. Values of M and K_a determined as a function of T without the elimination of $\dot{\varepsilon}$ are shown in Figs 3 and 4. It is also found that the Lüders strain ε_L is rate sensitive as it is shown in Fig. 5.

4. Discussion and conclusions

As seen from the results, Figs 2–4, the factor $\alpha(T)$ characterizing dislocation/obstacle interaction deviates from the constantly diminishing curve and at $-$ 470 K it shows a maximum indicating the existence of dislocations in the aged state. The maximum of α also seems to indicate a small drag effect. Changes in the coefficient of dislocation multiplication M are also consistent. At $-$ 380 K a maximum occurs for two lowest strain rates which indicates an increase in the strain hardening rate at this temperature. Since the steel tested was relatively calm the maximum is not very high. However, it must be remembered that even small changes in the coefficient of dislocation multiplication M cause relatively large changes in the flow stress σ (the accumulation of strain hardening). At higher temperatures than $\sim$ 600 K where Lüders strain vanishes, a substantial increase of M occurs. The annihilation factor K_a, remains fairly constant within the temperature range $300\,K < T < 600\,K$ and next at higher temperatures than 600 K shows a negative rate sensitivity as expected [6, 7]. The lack of rate sensitivity of K_a at the lower range of T, and its overall lower values to those expected, indicates the presence of the dynamic ageing. Within the dynamic ageing region the constancy of K_a is accompanied by an increase of strain hardening rate θ and by an increase of M. The constancy of K_a and maximum of M must be related to the instantaneous relative value of the concentration of interstitial atoms $\Delta C/C_0$ which diffuse to the immobile dislocations. At present there are no plausible quantitative models of plastic flow which take into account dynamic strain ageing. Results like those described above and others [1, 2, 11–13] may constitute the basis for improvements in modelling, for example Equation 2. Such a project is in progress.

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