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Influence of stacking-fault energy on high temperature creep of alpha titanium alloys

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Abstract

The stacking-fault energy (SFE) has been incorporated in the calculation of the steady-state creep rate of commercially-pure titanium and Ti–(1–10) mol.% Al alloys. Their creep behaviour was found to follow power-law creep when the dependence of SFE on the aluminium content was taken into account in the calculation. The possible contribution due to the formation of short range ordered structure is also discussed, with the conclusion that the decrease in SFE with increasing aluminium addition is the major factor for the strengthening effect in α Ti–Al alloys.
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The high temperature creep strength of a series of Ti–Al alloys has recently been studied by Oikawa and co-workers [1,2]. A continuous increase in the creep strength up to a concentration of 5 at.% Al was observed, whereas a Ti–7Al alloy showed an increase in creep strength larger than that expected from the increasing trend in the range Ti–(1–5)Al. Koike et al. [3] later investigated the tensile behaviour of Ti–(7–10)Al at 1100 K at various strain rates and proposed the following equation to analyse their results:

$$\dot{\varepsilon}_{\mathrm{s}} = \frac{AGb}{kT} N_{\mathrm{Al}}^{m} \left(\frac{\sigma}{G}\right)^{n} \exp\left(-\frac{Q}{RT}\right) \tag{1}$$

where A is a material constant, G the shear modulus, b the Burgers vector, N_{Al} the mole fraction of Al, m the concentration exponent, σ the applied stress, n the stress exponent, and Q the activation energy. RT and kT have the usual meanings.

Best fit values for m and n were determined by Koike as lying in the ranges -1.3 to -2.3 and 3.5–4.6, respectively. Whereas this range of values for n seems to be in line with power-law creep [4], the fact that m had to be given negative values seriously jeopardises the application of Eq. (1) to commercially pure (CP) titanium, because the N_{Al}^{m} term approaches infinity when the Al content approaches zero.

Since the concentration term N_{Al}^{m} was introduced solely to account for the unexpected strengthening effects due to the addition of aluminium, it would be preferable to replace this arbitrary parameter with a more physically realistic function for the strengthening effect.

The stacking-fault energy (SFE) is known to affect the deformation behaviour of metals and its influence on creep was firstly considered by McLean [5] in 1962. He suggested that the amount of primary creep may be a strong function of SFE. Weertman [6] later pointed out that a strong SFE effect on the steady-state creep rate could be expected in high temperature creep controlled by dislocation climb. Barrett and Sherby [7] empirically studied the influence of SFE on the creep of pure metals with the face-centred cubic (fcc) structure and found the steady-state creep rate, $\dot{\varepsilon}_{\mathrm{s}}$, to be proportional to $\gamma^{3.5}$, with γ denoting the value of SFE. For dimensional consistency, this is now more generally written in the form

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$$\dot{\varepsilon}_s \propto (\gamma/Gb)^p$$

where G and b are the shear modulus and the Burgers vector, respectively. However, different values for p have been proposed by Mohamed and Langdon [8] (1.2–4.2), Argon and Moffatt [9] (2), and by Yang et al. [10,11] (3). The value of p will be discussed later in the context of this paper.

If SFE is to be used as a parameter in a revised creep equation, the required values must either be experimentally available or be capable of being calculated for alpha titanium alloys.

The experimental SFE is reported to be is 0.31 J/m^2 for pure titanium, with thermochemical and first principle calculations providing values in the range 0.30–0.35 J/m^2 [12]. Thermochemical calculation relates the SFE of pure titanium to the difference in Gibbs energy between the fcc and hexagonal-close-packed (hcp) structures. The quoted first principle calculations were restricted to calculating basal plane faults and more recent atomic-scale modelling work indicates they are very model dependent [13]. No calculations seem to have previously been reported for titanium alloys, but for Ti–Al alloys it is possible to derive values via measured stacking fault probabilities (SFP) as the SFE varies inversely with the stacking-fault probability (SFP) [14]. Metzbower [15] proposed the following empirical relationship between SFP, denoted by α, and the aluminium concentration N_{Al}:

$$\alpha = 0.003\exp(0.133N_{Al}) \tag{2}$$

It is implied that the SFE is being referred to the basal planes, although the determination of SFP includes other fault-affected X-ray reflections. A similar relationship has been previously shown to exist in solid solution fcc alloys [16]. The fact that such a relationship holds in solid solution hcp alloy is not surprising, since a stacking fault in the hcp phase represents layers of the fcc phase and vice versa. Although pure titanium (with c/a ratio being lower than the ideal ratio) is known to deform primarily by $\{1\bar{1}00\}$ prismatic slip rather than the (0001) basal slip at room temperature, Williams et al. found that the slip mechanism did change both with increasing temperature and the addition of Al [17]. The effect of increasing temperature was seen to suppress twinning and promote basal slip. Although the increasing amount of Al in titanium increases the critical resolved shear stress (CRSS) on both the prismatic and basal planes, the effect on the basal planes is not as significant as that on the prismatic planes. Slip on basal planes is therefore favoured when the amount of Al in titanium increases. Williams et al. stated that prism and basal slip should be about equally important at 300 K when the Al concentration reaches 5–6 wt.% (8.5–10.2 at.%), while the effect of temperature is such that the CRSS for both slip modes is virtually identical in the range 400–1000 K for a 6.6 wt.% (11.1 at.%) alloy. Considering the temperature range for all the alloys studied and the addition of Al in many of the alloys, it is likely that a substantial amount of creep in these alloys involves dislocation movement related to the basal planes. It is therefore reasonable to propose that the effect of SFE on the creep properties of alpha titanium alloys will be similar to the established effect in fcc alloys.

Table 1
SFE values for various aluminium additions in alpha titanium alloys

Al (mol.%)	0	1	3	5	7	9	10
SFE (J/m^2)	0.310	0.271	0.208	0.159	0.122	0.094	0.082

When there is no aluminium addition, $\alpha = 0.003$ and the corresponding γ is 0.31 J/m^2. Assuming that the inverse relationship between α and γ holds, $\gamma = 0.093/\alpha$ J/m^2. Based on this relation, the SFE values for Ti–1, 3, 5, 7, 9 and 10Al alloys were calculated and are shown in Table 1. Studies by Karmaker et al. [18] on a Ti–Al–Mo–Zr alloy also confirmed that alloying of titanium with Al increased its SFP, i.e. decreased the SFE. The present work used the following equation to calculate the steady-state creep rate, which has previously been applied to nickel superalloys [19]

$$\dot{\varepsilon}_s = AD\left(\frac{\gamma}{Gb}\right)^3\left(\frac{\sigma}{E}\right)^4 \tag{3}$$

where A is a material constant, D is the diffusion coefficient, and E is the Young's modulus at the creep temperature. The calculation of D follows the usual expression: $D = D_0\exp(-Q/RT)$, where D_0 and Q are the pre-exponential term and the activation energy of the diffusion process, respectively. In this equation, the values of G, E, and D can be self-consistently calculated using computer software JMatPro® [20], which has been developed for the calculation of materials properties for the different phases present in multi-component alloys. The accuracy of JMatPro in calculating such properties has been extensively validated [21]. This leaves A as the only fitting parameter. The experimental data used in the present work was drawn from Refs. [1–3] for Ti–Al alloys and Refs. [22–25] for CP-Ti alloys, Table 2. The calculated creep rates are plotted against the experimental values in Fig. 1. It can be seen that Eq. (3) achieves a very good agreement between the experimental data and the calculated values.

In the above calculation, the exponent for the SFE function was set at 3, as previously used for fcc nickel based superalloys [19]. Other values in the range 1–3.5 did not give a better fit. When comparing Eqs. (1) and (3), it is clear that the former has an additional pre-exponential term which depends on $1/T$. However, as the temperature range of the reported experimental data is 573–1150 K, the maximum effect from this term on the calculated creep rate is only a factor of two. Attempts to include a temperature term $(1/T)$ into Eq. (3) resulted in a worse fit than Eq. (3) in its present form.

Ideally the SFE values that should be used in Eq. (3) are at the creep temperatures, but no experimental values are available. If one accepts that the SFE can be calculated from the Gibbs energy difference between the hcp and fcc

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