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The influence of stacking fault energy on the creep behaviour of Ni-Cu-solid-solution alloys at intermediate temperatures

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The creep characteristics of Ni–Cu alloys at intermediate temperatures ($T < 0.55 T_m$, where T_m is the absolute melting temperature), including the stress exponent (≥ 7) and the activation energy for creep (which is less than the activation energy for lattice diffusion), suggest that the creep mechanism is dislocation climb controlled by pipe diffusion. The present analysis shows that the creep rates of these alloys are consistent with a rate equation of the form

$$\dot{\varepsilon} = 50A\frac{D_p G b}{kT}(\Gamma/Gb)^3(\sigma/G)^7$$

where A is a dimensionless constant with a value of $\sim 10^{13}$, D_p is the pipe diffusion coefficient, G is the shear modulus, b is the magnitude of the Burgers vector, kT is the Boltzmann's constant times the absolute temperature, Γ is the stacking fault energy and σ is the applied stress. The Γ-values used in the present investigation were determined using high-temperature, lattice-diffusion, dislocation-climb-controlled creep rates. In addition, this equation can satisfactorily predict the pipe-diffusion-controlled creep behaviour in pure metals at intermediate temperatures.

1. Introduction

At high temperatures (above 0.6 T_m, where T_m is the absolute melting temperature), it has been established that the creep of pure metals and metal-class solid-solution alloys is controlled by the climb of edge dislocations. The activation energies measured in pure metals were in good agreement with the activation energies for lattice diffusion, suggesting that the climb process is diffusion controlled [1–3]. Under steady-state conditions, the creep rate, $\dot{\varepsilon}$, in pure metals and metal-class alloys [4–6] is usually written in the form

$$\dot{\varepsilon} = A\frac{D_c G b}{kT}\left(\frac{\Gamma}{Gb}\right)^3\left(\frac{\sigma}{G}\right)^5 \tag{1}$$

where G is the shear modulus, b is the Burgers vector, Γ is the stacking-fault energy, σ is the applied stress, k is the Boltzmann constant, T is the absolute temperature, A is a dimensionless constant and D_c is the climb diffusivity (which is equal to the self-diffusion coefficient in pure metals). However, in the case of solid-solution alloys, more than one species takes part in the diffusion process. It has been suggested [2, 7] that the Herring diffusivity [8], D_H, derived for diffusional creep in binary alloys, can describe the climb process in metal-class alloys; that is, D_c is equal to

$$D_H = \frac{D_A^* D_B^*}{(X_A D_B^* + X_B D_A^*)} \tag{2}$$

where D_A^* and D_B^* are the tracer diffusion coefficients of the solvent and the solute, respectively, X_A and X_B are the respective atomic fractions in the alloy A–B.

Recently theoretical analysis [9] and experimental discussions [10–13], have suggested that the following diffusion coefficient is more appropriate than the Herring diffusivity (given by Equation 2) for describing the climb diffusion coefficient; that is,

$$D_c = \frac{1}{f_0}(X_A D_A^* + X_B D_B^*) \tag{3}$$

where f_0 is a correlation factor close to unity.

At intermediate temperatures ($T < 0.6\ T_m$), the activation energies for creep in pure metals are lower than the activation energies for self-diffusion [14–20]. It is argued that pipe diffusion along the dislocation core [18–23] or cross-slip [24] may be the rate-controlling process. Recent experiments on pure Al [18], pure Ni [17, 19] and pure tin [20] support the suggestion that pipe-diffusion dislocation climb is rate controlling at intermediate temperatures.

To explain the lower activation energy and the high stress exponent (~ 7) observed at intermediate temperatures, an effective diffusion coefficient, D_{eff}, for climb [16, 21] has been used with Equation 1. The use of D_{eff}, which incorporates the additive contribution of the lattice and the dislocation pipe diffusion can change the stress exponent by a factor of 2 as a consequence of an increase in the dislocation density with an increase in stress [25, 26].

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Recently, Choi *et al.* [27] have studied the creep behaviour of Ni–Cu alloys at intermediate temperatures below 0.55 T_m. Based on the creep characteristics of these alloys including the activation energy for creep (which is less than the activation energy for lattice diffusion) and the stress exponent (≥ 7), Choi *et al.* [27] suggested that the dislocation pipe diffusion is the rate-controlling process. However, they did not attempt to make a quantitative correlation between their experimental results and the creep relations.

The results of Choi *et al.* [27] have motivated the present investigation. In the present paper, the D_{eff} approach will be used to correlate the steady-state creep rates in Ni–Cu alloys with a newly developed relation based on Equation 1. In addition, the measured creep rates in pure metals under the pipe-diffusion control conditions will be compared with the predictions of the new equation.

2. Analysis and discussion

The effective diffusion coefficient [16, 21, 28] is defined as

$$D_{eff} = (D_l f_l + D_p f_p) \tag{4}$$

where D_l and D_p are the lattice and pipe-diffusion coefficients, respectively, and f_l and f_p are the fractions of atom sites associated with the relevant diffusion process. The value of f_l is essentially unity, and $f_p = a_c\rho$, where a_c is the cross-sectional area of the dislocation core associated with rapid diffusion and ρ is the dislocation density. Taking $a_c = 5b^2$ and $\rho = 10/b^2(\sigma/G)^2$ and substituting in Equation 4, the following relation for D_{eff} [28] can be obtained

$$D_{eff} = \left[D_l + 50\left(\frac{\sigma}{G}\right)^2 D_p\right] \tag{5}$$

At high temperatures and low stresses, the lattice diffusion is dominant and the creep rate is represented by Equation 1. However, at low temperatures and/or high stresses the term $50(\sigma/G)^2 D_p \gg D_l$, and substituting D_{eff} for D_c in Equation 1 allows the following equation to be obtained for the pipe-diffusion-controlled creep.

$$\dot{\varepsilon} = 50A\frac{D_p Gb}{kT}\left(\frac{\Gamma}{Gb}\right)^3\left(\frac{\sigma}{G}\right)^7 \tag{6}$$

In order to compare the experimental creep rates in Ni–Cu alloys at intermediate temperatures with the predictions of Equation 6, estimates for the constant A and for the alloy parameters (such as G, D_p and (Γ/Gb)) are required.

The value of A can be determined by plotting the creep data for pure metals [2, 29] and for metal-class alloys [4, 6] at high temperatures in the normalized form of $(\dot{\varepsilon}kT/D_c Gb)/(\sigma/G)^5$ versus (Γ/Gb), using a double logarithmic scale, as shown in Fig. 1. Linear regression analysis of these data in Fig. 1, shows that the constant A in Equation 1 has a value of $\sim 10^{13}$ and the line of best fit has a slope of 3. This cubic dependence of creep rate on (Γ/Gb) is in agreement with a previous analysis [4]. This value of A is slightly higher than the value determined previously ($\sim 6.5 \times 10^{12}$) for metal-class alloys [6].

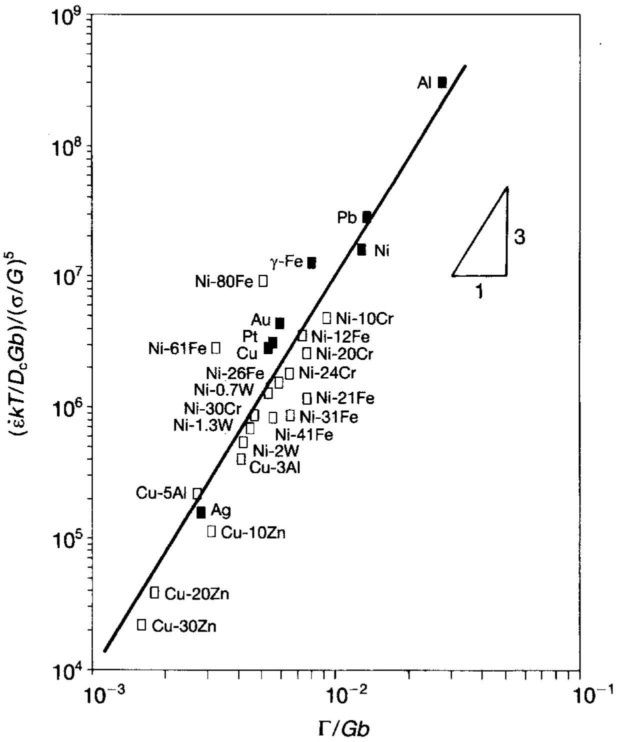

Figure 1 The stacking fault energy dependence of normalized creep rates of a number of face-centred cubic (f.c.c.) pure metals and metal-class solid-solution alloys.

The data for the stacking-fault energy in Ni–Cu alloys are very limited [30], and creep data for these alloys at high temperatures [31, 32] will be used to calculate (Γ/Gb) using Equation 1. The normalized creep rates for Ni–Cu alloys are calculated using D_c (from Equation 3) and they are plotted against the normalized stress (σ/G) on a double logarithmic scale in Fig. 2. The diffusion and shear-modulus data are taken from [2]; the G-values for the alloys were calculated using the mixture rule $(G = X_{cu}G_{Cu} + X_{Ni}G_{Ni})$. In addition, previously calculated normalized creep data [33], for Cu–Zn alloys [34] and Ni–W alloys [35] are included for purposes of comparison; the Γ-values for these alloys are available and their data correlate well with Equation 1 (see Fig. 1). As shown in Fig. 2, Cu–30Zn alloy (which has the lowest stacking-fault energy (17 mJ m^{-2}) exhibits the lowest creep rate for the same stress.

Using the normalized creep rates in Fig. 2 and using Equation 1 with $A \approx 10^{13}$, T/Gb values were calculated for the Ni–Cu alloys. The values of Γ which are plotted in Fig. 3 were calculated using the G-value of the alloy at room temperature since it is usually assumed that the parameter Γ/Gb is independent of the temperature [2, 4, 36]. The values calculated for Γ are lower than those recently determined by the rolling-texture technique [30]. As can be seen in Fig. 3, addition of Ni to Cu slightly decreases Γ from a value of 55 mJ m^{-2} for pure Cu to a value of 33 mJ m^{-2} for Cu–50%Zn. On the other hand, the addition of Cu to Ni sharply decrease Γ from a value of 250 mJ m^{-2} for pure Ni to 120 mJ m^{-2} for Ni–15%Cu.

Measurements for D_p in Ni–Cu alloys are not available. However, D_p is generally assumed to be equal to

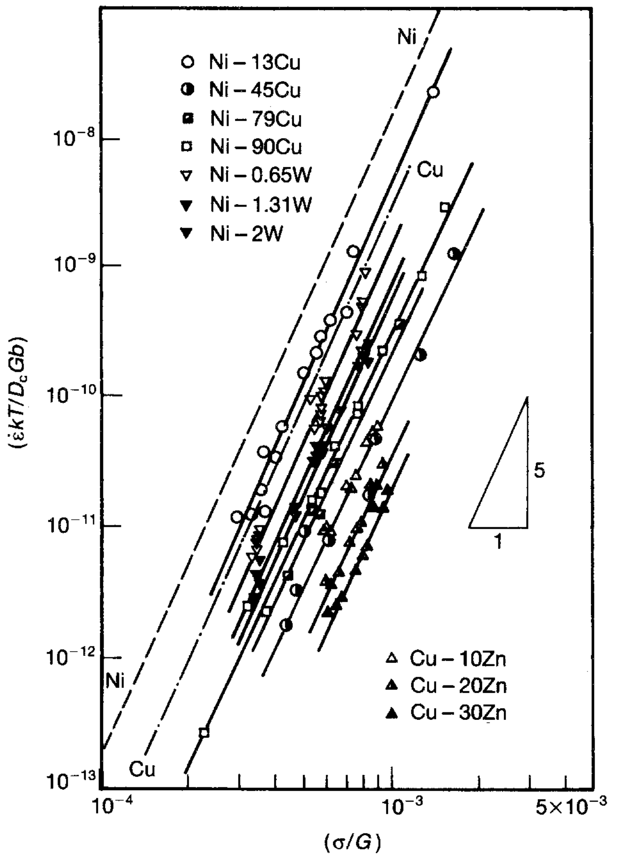

Figure 2 The normalized creep rate $(\dot{\varepsilon}kT/D_c Gb)$ plotted against the normalized stress (σ/G) for Ni–Cu alloys and for other metal-class alloys. Data for pure Ni and Cu are also included[2].

D_b, the grain-boundary diffusion [28]. The experimental data on grain-boundary diffusion have been reviewed recently [37], and it was found that Q_b depends on the temperature range; that is $Q_b = 9.35\,RT_m$ for $T_m/T < 2.4$ and $6.93RT_m$ for $T_m/T > 2.4$. It was very difficult to decide which value to be used because some of the creep-rate measurements in Cu–Ni alloys were taken very close to the boundary ($T_m/T \sim 2.4$). Fortunately, the creep activation energy, Q, was measured for these alloys [27]. For example, for Cu–15%Ni, Q was found to be 155 kJ mol^{-1} (37.2 kcal mol^{-1}), a value which is lower than the activation energy for creep at high temperatures (Q = 215 kJ mol^{-1}) [12] for the same composition. However, this measured value of Q for the Cu-15%Ni alloy is still higher than the Q_b-values mentioned above. The data for this alloy was chosen because of its large grain size, while the grain size of the other alloys which were tested was less than 40 μm and the creep rates of these alloys were grain-size dependent. For such a small grain size, the effect of the grain-boundary sliding on the measured creep rates was not negligible, and the activation energies calculated using these data are ambiguous and they cannot be used to represent the pipe-diffusion-controlled creep.

The creep rates for the Cu-15%Ni alloy at different temperatures are replotted in Fig. 4. As can be seen, Q decreases from a value of 227 kJ mol^{-1} (54.1 kcal mol^{-1}) at high temperatures to a value of 97 kJ mol^{-1} (23.1 kcal mol^{-1}) at low temperatures. The value for Q at high temperatures is very close to that deduced (~ 215 kJ mol^{-1}) [12] from the data of

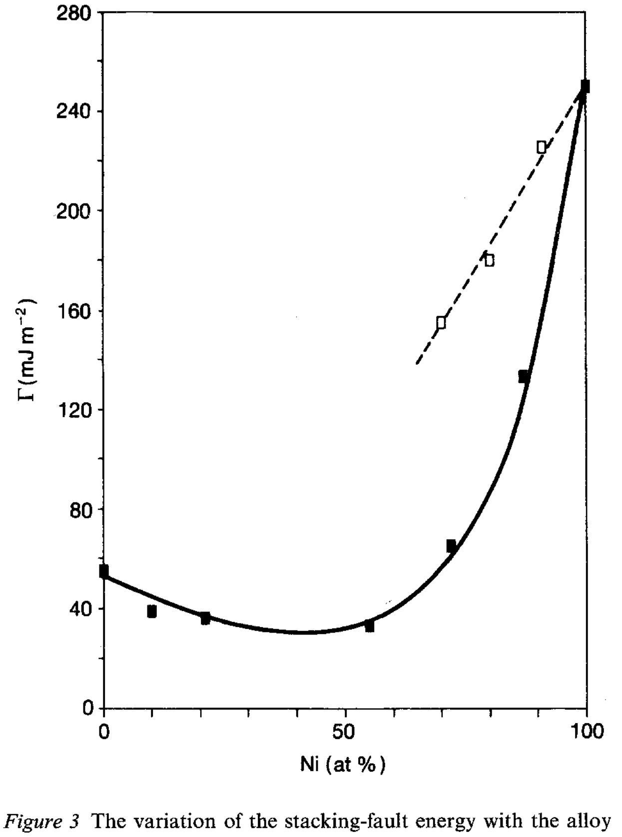

Figure 3 The variation of the stacking-fault energy with the alloy composition in the Ni–Cu system. (□) Ref. [30], (■) present.

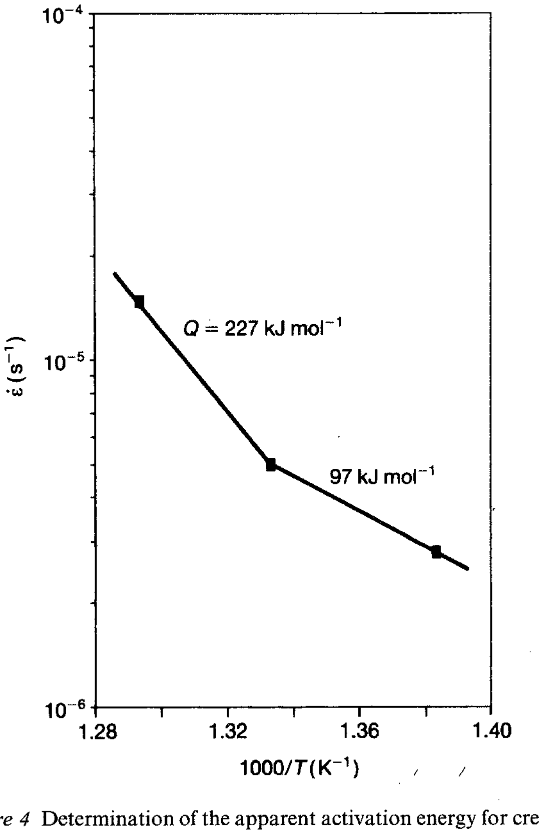

Figure 4 Determination of the apparent activation energy for creep in the Cu–15%Ni alloy by plotting log $\dot{\varepsilon}$ against $(1/T)$ at σ = 17.2 MPa.

TABLE I The predicted and experimental creep rates for Ni–Cu-solid-solution alloys tested at 17.2 MPa and 500 °C

Alloy (at %)	T/T_m	Grain size (μm)	Γ/Gb ($\times 10^{-3}$)	Q_p (kJ mol^{-1})	$\dot{\varepsilon}_{exp}$ (s^{-1})	$\dot{\varepsilon}_{cal}$ (s^{-1})[a]
Ni–16 Cu	0.46	27	6.78	114	7.5×10^{-8}	7.7×10^{-8}
Ni–31 Cu	0.48	20	3.40	109	4.9×10^{-8}	3.4×10^{-8}
Ni–48 Cu	0.50	24	2.18	105	1.0×10^{-7}	2.9×10^{-8}
Ni–57 Cu	0.51	21	2.37	102	1.7×10^{-7}	7.7×10^{-8}
Ni–71 Cu	0.52	31	2.63	100	8.1×10^{-7}	2.8×10^{-7}
Ni–85 Cu	0.54	53	3.13	97	1.48×10^{-5}	1.4×10^{-6}

[a] The creep rate calculated using Equation 6.

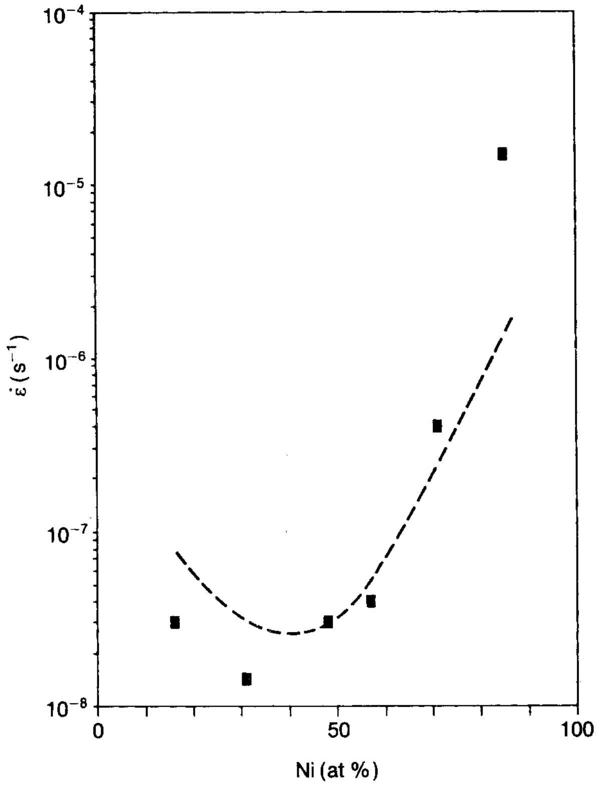

Figure 5 A comparison between (– – –) the creep rates predicted using Equation 6 and (■) the experimental grain-size-independent creep rates in Ni–Cu alloys.

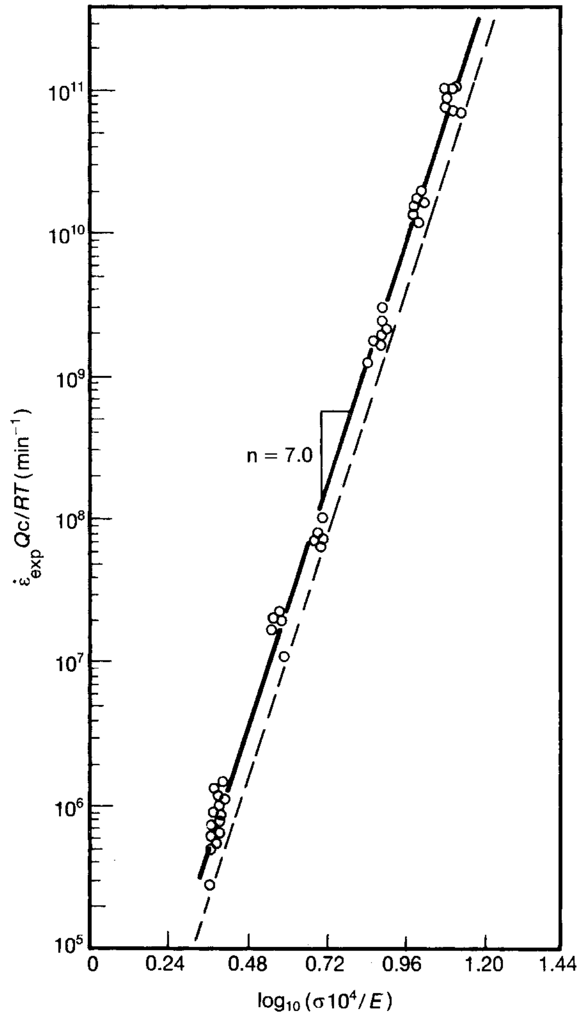

Figure 6 A comparison between (○) the experimental creep rates of pure Ni [17], and (– – –) the prediction of Equation 6.

Monma *et al.* [31] for Ni–Cu alloys at higher temperatures (1000–1273 *K*). The lower value for Q (97 kJ mol^{-1}) for Cu–15%Ni is comparable with the Q_b values; $Q = 8.2\,RT_m$, where T_m is the solidus temperature for the alloy in the phase diagram. This value for Q, which will be taken as a measure for Q_p, in Ni–Cu alloys, is an average value for the two values reported for Q_b. The pipe diffusion coefficient is taken as

$$D_p \;=\; 1\times10^{-4}\exp(-8.2\,T_m/T)\,\mathrm{m^2\,s^{-1}}$$

This value is not too far from those tabulated for the grain-boundary diffusion in f.c.c. metals and alloys [38].

The calculated values for $\dot{\varepsilon}$, based on Equation 6 using the estimated values for A, D_p and Γ/Gb are listed in Table I along with the experimental results. The agreement between the predictions of Equation 6 and the experimental data is not unsatisfactory considering the approximations introduced in estimating the values of D_p and the other parameters. The data point for Cu–15%Ni at 773 K is not well within the pipe-diffusion-control region; better agreement is obtained between the prediction of Equation 6 and the creep rates at lower temperatures. For example, at $T = 723$ K ($\sim 0.5\,T_m$), the predicted creep rate is $4.5\times10^{-7}\,\mathrm{s^{-1}}$ which is a factor of 6 below the experimental creep rate.

The grain-size independent creep rates for Ni–Cu alloys determined by Choi *et al.* [27] for specimens with grain size >40 µm are plotted in Fig. 5, along with the creep rates calculated using Equation 6 (Table I). The agreement between the experimental creep rates and the predictions of Equation 6 is more than satisfactory.

Finally, a comparison between the predictions of Equation 6 and the experimental pipe-diffusion-controlled creep rates for five f.c.c. pure metals at $T \approx 0.5T_m$ has been carried out and published elsewhere [39]. This comparison showed that Equation 6 can satisfactorily predict the creep rate in pure metals at intermediate temperatures. However, the creep behaviour of pure Ni [17], which was investigated over a wide range of temperatures (0.31–$0.55\ T_m$) and stresses ($6 \times 10^{-4} - 3 \times 10^{-3} G$), will be examined in the present paper for the sake of completeness. For pure Ni [17], the creep parameters, the stress exponent and the activation energy for creep, Q_c, are independent of the stress and of temperature over the stress and temperature ranges studied, and they have values of 7 and 41 kcal mol^{-1} (171 kJ/mol^{-1}), respectively; this is in qualitative agreement with Equation 6. In addition, the creep rates calculated using Equation 6 are compared with the experimental creep rates of Ni at various stresses and temperatures in Fig. 6. As can be seen, the calculated creep rates are only a factor of 3 slower than the experimental rates; the agreement is more than satisfactory.

3. Conclusions

1. The high-temperature creep rate of pure metals and metal-class alloys can be well represented by the equation

$$\dot{\varepsilon} = 10^{13} \frac{D_c G b}{kT} \left(\frac{\Gamma}{Gb}\right)^3 \left(\frac{\sigma}{G}\right)^5$$

2. The calculated values of the stacking-fault energy, Γ, for Ni–Cu alloys suggest that the addition of Ni to Cu slightly decreases Γ, while the addition of Cu to Ni sharply reduces the value of Γ.

3. The present analysis suggests, for the first time, that pipe-diffusion-dislocation-climb-controlled creep is similar at intermediate temperatures to high-temperature creep, and that it decreases with decreasing Γ and the functional dependence can be represented as $\dot{\varepsilon}\ \alpha (\Gamma/Gb)^3$.

4. The grain-size-independent creep rates of Ni–Cu alloys at intermediate temperatures can be satisfactorily predicted using a creep power law Equation 6 when incorporating the appropriate values for the alloy parameters such as the pipe diffusion coefficient, the stacking fault energy and the shear modulus.

5. Good agreement is obtained between Equation 6 and the measured creep rates in pure metals (such as Ni) at intermediate temperatures.

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