

Dislocation creep in particle-strengthened systems

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A model for climb-controlled creep in two-phase materials is proposed which invokes the presence of a back-stress for providing the necessary mobile dislocations for creep. It is shown that by incorporating this stress in the creep equation it is possible to reduce both the activation energy for creep and the stress exponents to values normally observed in single-phase materials. Creep data on TD nickel and yttriated superalloy when analysed on this basis confirm the applicability of this model.

1. Introduction

High-temperature alloys are usually particle-strengthened to provide for a high creep strength. Creep experiments on these alloys at temperatures where dislocation climb is predominant have shown that both the activation energy for creep and the stress exponent are high compared to pure metals and solid solutions. For instance, in nickel-base superalloys the activation energy has been measured to be around 600 to 800 KJ Mol⁻¹; the stress exponent could be as high as forty. A number of possible dislocation mechanisms have been suggested to account for these high values. Malu and Tien [1] have postulated that the rate-controlling mechanism for particle-strengthened systems could be the climb of interstitial-emitting jogs. Interstitial emission, according to them, becomes necessary because of a very fine dislocation substructure stabilized by particles. Lund and Nix [2] have pointed out some of the shortcomings of this theory and suggested the incorporation of temperature variation of shear modulus in the creep equation. This correction would by itself bring down the activation energy to that for self-diffusion. Malu and Tien [3] followed this procedure and applied the necessary shear modulus correction to creep data on Inconel Alloy MA-753*, but found that at lower temperatures these corrections were clearly inadequate to bring down the activation energy. They went on to point out

that the mechanism originally suggested by them would, on the other hand, account for the high activation energy even at these temperatures.

Recently there have been a few attempts to invoke the concept of a back stress to explain the creep behaviour of these materials. Pattanaik *et al.* [4] have shown that by subtracting a back-stress equal to the stress necessary for a dislocation to climb over a particle from the applied stress, the stress exponent in the Dorn Equation could be reduced to about three. Such an exponent would bring the creep behaviour of particle-strengthened systems in line with that of single-phase materials [5]. Small n values could also be interpreted as having some physical significance [6]. Though the concept of invoking a back-stress in explaining the creep behaviour is not new, its applicability to creep in two-phase systems and the physical processes involved have not until now been made clear.

This paper describes climb-controlled creep in particle-strengthened systems invoking the presence of a back-stress. It is shown that by incorporation of a back-stress in the creep equation, it is possible to reduce both the activation energy and the stress exponent to values observed in single-phase systems. The back-stress is shown to have a physical significance associated with dislocations overcoming the particles.

* A registered trade name of the International Nickel Company.

2. Back stress in climb-controlled creep

For climb-controlled creep to occur in particle-strengthened systems, it is first necessary to provide sufficient stress to dislocations for overcoming the particles by one of the following routes: climbing over the particles, looping around them, or cutting through them. All of these processes involve a significant amount of energy dissipation, which can be large, localized and irrecoverable, making an externally applied stress necessary. Until this stress is forthcoming, dislocations will be unable to free themselves from obstacles. Creep, then, cannot occur except by diffusion. The applied stress can therefore be thought of as opposing a "back-stress", and the effective stress is the difference between the applied stress and the back-stress. The effective stress necessary for dislocations to climb over thermally penetrable obstacles can be the same in particle-strengthened as well as particle-free matrices. This is true not only in creep but also in tensile deformation. Tensile experiments in thoriated nickel and nichrome have shown that until the necessary back-stress is provided for, there is no effective stress, and hence no strain-rate sensitivity [7].

When dislocations climb locally over particles they have to acquire extra line length to cater to their intermediate configurations. Brown and Ham [8] have shown that the stress necessary for climb over cuboidal particles is constant and equal to $\sigma_{\text{Orowan}}/\sqrt{2}$. Similar order-of-magnitude calculations for spherical particles [4] have shown a back-stress of $0.57 \sigma_{\text{Orowan}}$. Lagneborg [9] has shown that the configuration assumed by Brown and Ham is restrictive in that climb is assumed to

be localized around the particle. A more realistic assumption would be for climb to extend over half the inter-particle spacing. This would smoothen out the configuration making climb gradual, and distributing the climb over the entire line length. In this model, the back-stress is not constant, but rises proportionally with applied stress up to a limiting value which is the Orowan stress for cutting. The variation of back-stress with applied stress is shown schematically in Fig. 1a.

When incoherent particles are present new interfaces are created, leading to dislocations relaxing at these interfaces. Extra energy must therefore be supplied to unpin them. Petrovic and Ebert [10] have estimated the back-stress for this process to be $2E_s/R_0$, where E_s is the energy of the interface and R_0 , the radius of the particles. This would contribute an extra constant term to the back-stress and until this stress is reached there would be no climb over the particles. The relation between the back-stress and applied stress is thus modified for the incoherent particles and is shown schematically in Fig. 1b. Here the back-stress is equal to the applied stress till $2E_s/R_0$ is reached. Above this stress the back-stress increases in proportion to the applied stress levelling off at the Orowan stress as in Lagneborg's model. When the interface energy is high, the Orowan stress itself could be the minimum stress required for overcoming the particles.

3. A modified creep equation

In the preceding section we introduced the concept of a back-stress in particle-strengthened systems and showed how this stress is related to the

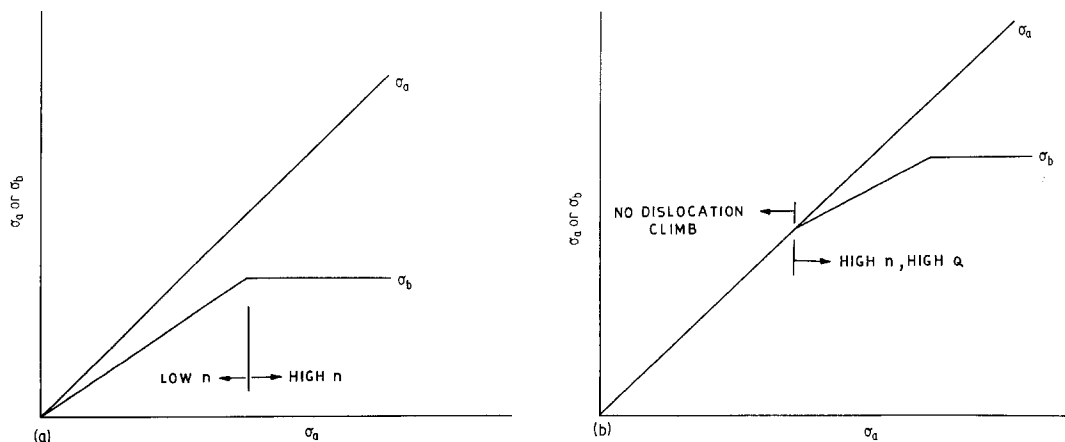


Figure 1 (a) Schematic representation of the dependence of back-stress σ_b , on applied stress σ_a , for precipitation-hardened alloys. (b) Schematic representation of the dependence of back-stress on applied stress for dispersion-strengthened alloys.

event of dislocations overcoming the particles. An applied stress equal to the back-stress does not contribute directly to climb-controlled creep except in providing free dislocations for climb, and must therefore be subtracted from the applied stress when formulating the creep equation. The creep equation should then read:

$$\dot{\epsilon} = A \left(\frac{\sigma_a - \sigma_b}{E} \right)^n \exp \left[- \frac{Q_{\text{real}}}{RT} \right], \quad (1)$$

$\dot{\epsilon}$ is the strain rate, A is a constant, σ_a is the applied stress, σ_b the back-stress, n the stress exponent, and Q_{real} is the real activation energy for creep. E , R and T have their usual meanings.

The true activation energy, Q_{real} , is less than the apparent activation energy Q_{app} , which is obtained when the creep equation does not take into account either the temperature variation of elastic modulus or the presence of a back-stress:

$$Q_{\text{real}} = Q_{\text{app}} + n \frac{RT^2}{E} \frac{dE}{dT} + n \frac{RT^2}{\sigma_a - \sigma_b} \frac{\partial \sigma_b}{\partial T}. \quad (2)$$

When these corrections are introduced, Q_{real} approaches the activation energy for self-diffusion. The stress exponent also reduces to a value close that observed in single-phase materials [5]. Lund and Nix [2] have suggested the incorporation of only the shear modulus term in the creep equation to reduce the activation energy.

Certain consequences of this modified equation can now be pointed out: (1) As the total applied stress in its entirety is not available for creep, but only a much lower stress, creep becomes subdued in systems containing second-phase particles. This decrease has been observed by Parker and Wilshire [11]. However, they have attributed it to the reduced growth rate of dislocation networks in the presence of particles. (2) For hard coherent particles at low applied stresses, where the back-stress is proportional to the applied stress, large n values would not be observed in single-phase materials. With increasing applied stress, the back-stress could reach saturation. Above this value the proportionality between σ_a and σ_b would be lost, leading to large n . This has indeed been observed experimentally in some nickel-base superalloys [12]. (3) For incoherent particles, where the back-stress is not proportional to the applied stress, large n would be seen at all stresses. n would not

be constant but would decrease gradually with increasing stress. At high stresses both n and Q should come down to values observed for solid solutions. The trend of a decreasing n has been observed in SAP by Hansen and Clauer [13]. Lund and Nix [14] have also observed similar behaviour in single crystals of Ni–20Cr–2ThO₂. They have interpreted this by invoking the Orowan loop criterion. (4) High n and high Q need not necessarily be coupled: High n arises whenever the back-stress is a significant fraction of the applied stress and relatively insensitive to changes in the applied stress. High Q is found when the backstress is large and varies with temperature. If the temperature dependence of σ_b is small, Q will be low, even though n can be high. Alternatively, if σ_b varies rapidly with temperature, Q will be high even when n is low. (5) For determining the activation energy for creep, or for evaluating the long-term creep behaviour it is usual to carry out accelerated creep tests by increasing either the applied stress or by changing the temperature during creep experiments. The presence of a variable functional dependence of back-stress on applied stress makes stress change experiments unreliable for extrapolation.

4. Comparison with experimental results

Benjamin and Cairns [15] have experimentally determined creep parameters in a multiphase nickel-base superalloy Inconel Alloy MA-753 containing both γ' and Y₂O₃. Results of these experiments have been analysed by Malu and Tien [3] who showed that correction due to shear modulus variation alone is insufficient to account for the high Q at lower temperatures.

We have re-examined the results of Benjamin and Cairns [15] after subtracting the back-stress from the applied stress. The back-stress values were calculated indirectly, assuming that n for the solid solution in the absence of particles would be near five. The stress to be subtracted from the applied stress to bring n from the observed high values to five was computed at various temperatures, and the temperature-dependence of this stress is shown in Fig. 2. Elastic modulus variations with temperature were also incorporated in the creep equation. Results of the calculations are shown in Table I. The activation energy for creep now comes down to about 350 kJ mol⁻¹ at all temperatures which is almost the activation energy for creep in nickel–chromium solid solutions. An

TABLE I Summary of results for dispersion strengthened superalloy (data of Benjamin and Cairns [15])

Temp. (K)	σ_a (MN m ⁻²)	Q_{app} (kJ mol ⁻¹)	n	$E \times 10^4$ (MN m ⁻²)	dE/dT (MN m ⁻² K ⁻¹)	σ_b (MN m ⁻²)	$\partial\sigma_b/\partial T$ (MN m ⁻² K ⁻¹)	Correction to Q due to modulus variation (kJ mol ⁻¹)	Correction to Q due to back-stress (kJ mol ⁻¹)	Q_{real} (kJ mol ⁻¹)
977	415	640	5	18.1	-89.5	259	-1.1	-12	-280	348
1200	158	810	5	15.4	-126	122	-0.24	-56	-392	362
1311	131	835	5	13.8	-176	88	-0.24	-90	-402	343

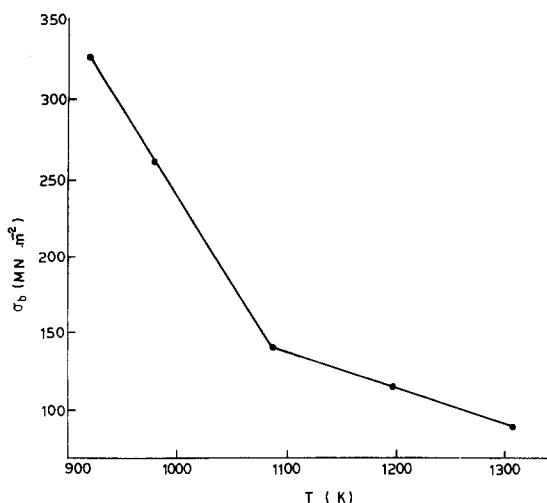


Figure 2 Temperature dependence of back-stress for yttria dispersed superalloy.

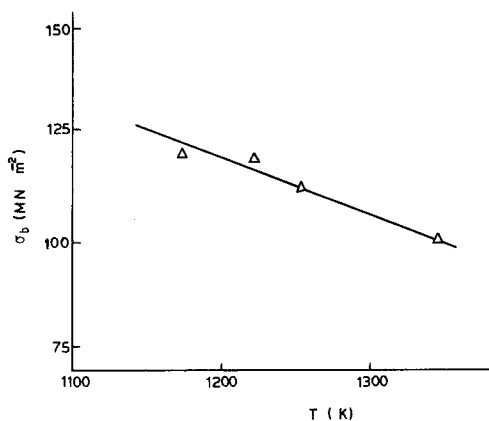


Figure 3 Temperature dependence of back-stress for TD nickel.

interesting feature in these calculations is the variation of back-stress with temperature (Fig. 2). The trend observed here is similar to that observed in the variation of yield stress with temperature for these dispersion-strengthened superalloys.

Experimental results on TD Nickel by Wilcox and Clauer [16] have also been analysed in the above fashion. Results of these analyses are shown in Table II and in Fig. 3. There is only a general agreement between Q_{real} and the activation energy for self-diffusion. Part of the difficulty is due to the back-stress being close to the applied stress, thus making the activation energy very sensitive to the choice of the back-stress.

Hansen and Clauer's results [13] on SAP yield a true activation energy of 190 kJ mol⁻¹. This was derived by assigning an n value of 4.5 to aluminium. Another observation of note in these results is the very rapid fall of creep rate with a small decrease in the applied stress, which would suggest presence of a threshold stress below which there would be no creep by climb. This fits in well with the hypothesis that a certain stress must first be provided for dislocations to free themselves from particles before they could creep.

5. Conclusions

High n and Q values observed during creep in two-phase materials is due to the athermal barrier offered by second phase particles to dislocation motion. Until this stress barrier is overcome there can be no free dislocations for climb.

TABLE II Results from modified formulation for TD nickel (data of Wilcox and Clauer [16])

Temp. (K)	Stress range (MN m ⁻²)	σ_a (MN m ⁻²)	σ_b (MN m ⁻²)	n	$\partial\sigma_b/\partial T$ (MN m ⁻² K ⁻¹)	Correction on Q due to modulus variation (kJ mol ⁻¹)	Correction on Q due to back-stress (kJ mol ⁻¹)	Q_{app} (kJ mol ⁻¹)	Q_{real} (kJ mol ⁻¹)
1171	124-138	131	115	5	-0.126	-50	-450	768	268
1218	124-138	131	114	5	-0.126	-54	-460	768	254
1250	117-124	120.5	107	5	-0.126	-63	-580	768	125
1343	103-117	110	96	5	-0.126	-70	-670	980	240

Incorporation of a back-stress in the creep equation brings out the true values of activation energy for creep and the stress exponent. These values are similar to the ones measured in single-phase materials, suggesting a universal validity of Dorn's equation to climb-controlled creep in all materials.

Application of the modified creep equation to results obtained on creep in various two-phase materials has confirmed the general applicability of this model.

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