Creep strength contribution due to precipitation hardening in copper–cobalt alloys

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ABSTRACT

In spite of its huge technical significance, there does not seem to be consensus about how to model the precipitation contribution to the creep strength. Most contributions in the literature are based on a constant internal stress (also called back stress or threshold stress) from the precipitation. It is well-known and it will also be demonstrated in the paper that this assumption is at variance with observations except for some ODS alloys. There is, however, one model developed by Eliasson et al. (Key Eng Mater 171–174:277–284, 2000) that seems to be able to represent experimental data without the use of any adjustable parameters. It has successfully been applied to describe the creep strength of austenitic stainless steels. Due to the fact that various mechanisms contribute to the creep strength in these steels, the model has not been fully verified. The purpose of this paper is to apply the model to published creep data for Cu–Co alloys, where the precipitation totally dominates the strength contribution to validate the model. In the paper, it is demonstrated that the model can indeed describe the influence of applied stress, alloy composition and heat treatment for the three analysed Cu–Co alloys.

Introduction

General methods to increase the resistance to plastic deformation of metallic materials are cold working, precipitation hardening and solid solution strengthening [1]. Precipitation hardening is often the most efficient way to improve the creep strength of high-temperature alloys. A classical approach to analyse precipitation hardening during creep is to introduce a constant internal stress σ_i in a power law creep rate expression [2].

$$\dot{\epsilon} = A(\sigma - \sigma_i)^n \quad (1)$$

where A is a constant at a given temperature and n stress exponent for the alloy corresponding to the matrix. The internal stress is often also called back stress or threshold stress. σ_i was in general treated as an adjustable parameter for fitting a linear relationship of double logarithmic creep rate versus reduced stress. A direct way of determining σ_i was obtained by using a Lagneborg–Bergman plot [3]. A constant σ_i results in a threshold stress below which no creep takes place. For oxide dispersion strengthening (ODS) alloys such a threshold has been observed

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[4, 5] and it can be essentially independent of temperature. Equation (1) gives a creep exponent that decreases with increasing stress. This has been found for some ODS alloys [5], but rarely for precipitation strengthened steels, where the stress exponent is raised with increasing stress. Consequently, a constant threshold stress is not applicable to precipitation strengthened alloys.

Orowan bowing has often been used to estimate the threshold stress but it usually results in an overestimate of the creep strength, since dislocations can climb across the particles at high temperatures. A threshold stress has also been determined by taking climb into account [6]. In the past, many attempts were made to compute this threshold stress and the predicted values were quite different, depending sensitively on the assumptions of the models [7]. Two kinds of models have been brought forward. One kind was based on local climb assumption proposed by Brown and Ham [8], which postulated that the climbing dislocation segment was attached to the particle and the dislocation between the particles remained in its glide plane. The predicted stress was about half of the Orowan stress, which was consistent with the experimental values of some precipitation strengthened materials. However, the assumption itself was unrealistic since the sharp bend in dislocation can be rapidly relaxed by diffusion. Under the assumption of general climb, the climbing dislocation is only in contact with the particle at a single point. Gradually researches found decreasing threshold stress [4, 5, 9]. Only threshold stresses as low as 0.02–0.06 of the Orowan stress were eventually predicted [4, 5]. The general climb assumption was more realistic than the local climb assumption and has been demonstrated to be the energetically favourable case. However, the stress or energy barrier for dislocation to climb over particles was quite small which cannot explain the contribution of particles to the creep strength in general [5, 7]. One possibility is to assume an attractive dislocation-particle interaction. Such an interaction has been observed for oxide dispersion strengthened (ODS) alloys [10, 11], but rarely for precipitation hardened systems. Another important difference between dispersion and precipitation strengthened systems is that a true threshold stress can be present in ODS alloys [5]. In such a case, the creep rate is negligible below the threshold stress. However, for precipitation strengthened materials, the situation is different. For example, for common

creep resistant CrMo steels that are precipitation strengthened, no threshold stress is found. There are also ODS alloys where no threshold stress is observed [6].

In general, the climb threshold stresses are too low to explain the observed influence of particles on the creep strength. To handle this situation, a time controlled climb model was then proposed by Eliasson et al. to evaluate the actual amount of climb [12]. For particles with different sizes, it will take different time for a dislocation to climb across. A critical size is introduced to describe the maximum particle radius that the dislocations can climb within the lifetime. The assumption of this model is that the key controlling mechanism for a dislocation to climb across a particle or not is the time it takes. The actual amount of climb is calculated by comparing the time for dislocations to glide and to climb. Only particles that are large enough cannot be passed by climbing dislocations give a contribution to the creep strength. The model has been used to evaluate the particle contribution for 9 and 12Cr steels and austenitic stainless steels [13, 14]. A good agreement with the experimental values was obtained.

In fact, creep strength is gained not only from precipitation hardening but also from solid solution hardening and dislocation densities. In previous studies, the creep strength has been calculated taking the three contributions into account. In the present paper, the contribution of precipitation hardening to the creep strength is evaluated for previously published creep data for copper-cobalt alloys with different cobalt content and different particle size distribution [15, 16]. It is verified that the contribution from solid solution hardening is negligible. The aim of the present paper is to study the precipitation hardening mechanism using copper-cobalt system without having to take into account the influence of many kinds of particles or of elements in solid solution. In this way, the accuracy of the precipitation hardening model can be verified.

Material and testing

Creep data for three copper-cobalt alloys (with 0.88 wt.%, 2.48 wt.% and 4.04 wt.% cobalt) published by Wilshire et al. are analysed in the present paper [15, 16]. Copper-cobalt is a good system for the study of precipitation hardening since it is a simple alloy

containing only coherent cobalt particles with little disturbance from elements in solid solution. The precipitates nucleate shortly after the ageing process starts, so the volume fraction of precipitates remains constant during the ageing process. The ageing was performed at a higher temperature than the creep test, and in this way the creep-induced precipitation could be avoided. In addition, the precipitation hardening is the dominant contribution to the creep strength.

The three copper–cobalt alloys were solution treated followed by water quenching, which resulted in a uniform grain size of about 200 nm. In order to obtain a uniform distribution of particles, ageing treatment was carried out for the alloys during various time. The detailed ageing procedures are given in Ref. [15]. Cu4.04Co was aged at a higher temperature than the other alloys due to sluggish precipitation, but the alloy was stabilized at the lower heat treatment temperature to avoid precipitation during creep. After ageing, the particle sizes were measured under microscope. The particle characteristics for the different alloys is summarized in Table 1.

Micro-hardness for the Cu0.88Co alloys at different ageing time was measured. The heat treatment that gave maximum micro-hardness is referred to as (fully) aged. When the radius is smaller than in the aged condition, it is called underaged and when it is larger overaged. The volume fraction of the precipitates and the amount of Co in solid solution were calculated with the thermodynamic software Thermo-Calc. The interparticle spacing was determined with the expressions for planar square lattice particle spacing λ_s ,

$$\lambda_s = \bar{r} \sqrt{\frac{2\pi}{3f}} \quad (2)$$

where $\bar{r}$ is the mean radius of the particles and f the volume fraction of precipitate. The constant stress tensile creep tests were conducted at 439 °C. No particle growth was recorded during the creep tests [15]. In the underaged condition, the strength is controlled by particle cutting, most likely by coherency shearing for the Cu–Co system and not by the Orowan stress. Unfortunately, the contribution from coherency shearing to the room temperature strength is difficult to estimate but it decreases with decreasing particle radius. This is the reason for the lower hardness in the underaged condition. Particle cutting does not contribute to the creep strength, since fine particles are easily climbed by the dislocations, see below.

Model for precipitation hardening

Dislocation creep

It will be assumed that the precipitation hardening during creep can be taken into account with the help of an internal stress. This assumption has practically always been made in the literature. The internal stress is also called back stress or threshold stress. When a threshold stress is introduced, it is in general assumed as a constant, i.e. is essentially independent of temperature and applied stress. This terminology will be followed in the present paper. But before the details about the precipitation hardening can be given, a model for secondary creep is needed for a

Table 1 Analysed Cu–Co alloys

Co (wt.%)	Heat treatment temperature (°C)	Heat treatment type	Particle volume fraction	Co in solid solution (wt.%)	Particle radius (nm)	Particle spacing (nm)	Hardness (HV)	Orowan stress (MPa)
0.88	600	Underaged	0.00567	0.33	1.2	28	85	(1031) ^b
0.88	600	Aged	0.00567	0.33	4.2	60	92	435
0.88	600	Overaged	0.00567	0.33	8.6	135	83	210
2.48	600	Aged	0.0222	0.33	7.6	60	99	473
4.04	700	Aged	0.0344	0.33*	22.6	215	93	197

*Stabilized at 600 °C after heat treatment at 700 °C

^bIn the underaged condition, the Orowan stress does not represent the strength

particle free material. In the past, the influence of applied stress, σ , and temperature T on the secondary creep rate $\dot{\epsilon}_{\text{sec}}$ has been represented by the conventional power law and Arrhenius equations in the form

$$\dot{\epsilon}_{\text{sec}} = A \frac{D_{0\text{sd}} G b}{k_B T} \left(\frac{\sigma}{G} \right)^n \exp \left(- \frac{Q_{\text{sd}}}{R_g T} \right) \quad (3)$$

where G is the shear modulus, b the length of the Burgers vector, k_B the Boltzmann's constant and R_g the gas constant. $D_{0\text{sd}}$ in Eq. (3) is the pre-exponential factor in the Arrhenius equation for self-diffusion and Q_{sd} is the activation energy for self-diffusion in the alloy. There are several limitations of Eq. (3). There are at least two constants, A and n , that are used as adjustable parameters and fitted to experimental data. This means that the equation cannot be used for predicting creep properties. This makes it also quite risky to extrapolate results to new conditions. Furthermore, its use is restricted to the power law regime where n is approximately constant, so it cannot describe creep at lower temperatures where the creep exponent n rapidly increases with decreasing temperature.

To avoid all these limitations, a new model for secondary creep has been developed. It was originally formulated for copper [17] but has been shown to be valid also for aluminium alloys [18, 19] and stainless steels [20, 21]. The model is derived from fundamental dislocation formulations. The derivation can be found in several places [22, 23] and it will not be repeated here. The important aspect is that it is based on a balance between work hardening and recovery in the secondary stage. The resulting expression for the secondary creep rate is

$$\dot{\epsilon}_{\text{sec}} = \frac{2\tau_L b c_L}{m} M_{\text{cl}}(T, \sigma_{\text{disl}}) \left(\frac{\sigma_{\text{disl}}}{\alpha_T m G b} \right)^3 \quad (4)$$

τ_L is the dislocation line tension, c_L a work hardening parameter, and $m = 3.06$ the Taylor factor. The constant α_T appears in Taylor's equation

$$\sigma_{\text{disl}} = \alpha_T m G b \rho^{1/2} = \sigma - \sigma_i \quad (5)$$

σ_{disl} is the dislocation stress. σ_i is an internal stress that can have contributions from the yield strength, solid solution hardening and precipitation hardening. The dislocation climb mobility is given by

$$M_{\text{cl}}(T, \sigma) = \frac{D_{0\text{sd}} b}{k_B T} e^{\frac{\sigma b^3}{k_B T}} e^{-\frac{Q_{\text{sd}}}{R_g T}} \left[1 - \left(\frac{\sigma}{R_m} \right)^2 \right] \quad (6)$$

where R_m is the tensile strength (at room temperature). All the parameters in Eqs. (4)–(6) are given and none is used as adjustable parameter and fitted to the mechanical test data. Equation (4) is generalized to describe primary and tertiary creep as well [23–25] but since only secondary creep will be discussed in the present paper, these extensions will not be considered here. When Eq. (6) is first developed, it is referred as a combined climb and glide mobility. The reason was that at high temperatures it agreed with the established climb mobility [26], but at near ambient temperatures, it was believed that the enhanced value from the last factor in the equation was due to glide [17]. However, recent research where the role of strain-induced vacancies is taken into account has demonstrated that Eq. (6) represents the climb mobility down to near ambient temperatures [18, 19]. Consequently, Eq. (6) is now referred to as the climb mobility.

Precipitation hardening

Three main mechanisms for how dislocations bypass particles are in general considered: particle shearing, Orowan bowing and climb across particles. The stress required for Orowan bowing σ_O is

$$\sigma_O = \frac{m C G b}{\lambda} \quad (7)$$

where λ is the interparticle spacing and $C = 0.8$ is a constant [27]. σ_O is only weakly temperature dependent (through G). Equation (7) has sometimes been used to estimate the particle contribution to the creep strength, but in general it will largely overestimate the contribution. This will be exemplified in Fig. 5. The reason is that the stress required for climb is much less. Initially, the threshold stress was estimated to be quite high, but later studies gave lower values [2, 4, 5, 8, 28]. Reviews are given in [5, 6]. The best estimate of the threshold stress, i.e. the minimum climb stress σ_{clmin} is now considered to be [4, 5]

$$\frac{\sigma_{\text{clmin}}}{\sigma_O} = \frac{\alpha}{\alpha + 2C} \quad (8)$$

where

$$\alpha = \frac{2\bar{r}}{3\lambda} = \sqrt{\frac{2f}{3\pi}} \quad (9)$$

The minimum climb stress is related to the Orowan stress through the parameter α , which is referred to as

the *climb resistance*. The final step in Eq. (9) is obtained by assuming that the planar square lattice particle spacing $\bar{r}\sqrt{2\pi/3f}$ is used for λ . α is quite small. For volume fractions f of 1 and 5%, α is 0.05 and 0.10, which gives threshold stresses of $0.03 \sigma_O$ and $0.06 \sigma_O$, respectively. Later analysis has confirmed these low values [7].

Although the threshold stress is quite small, it is not negligible but it obviously does not give the full picture. To handle this situation, a different track will be followed. The following assumptions will be made [12, 13].

- Only precipitation hardened alloys will be considered. Consequently, the attractive interaction between particles and dislocations will be neglected.
- It will be assumed that it is the time it takes for a dislocation to climb across a particle that is the controlling mechanism instead of the threshold stress. Thus, it will be the time it takes for the dislocation to climb across the particles that determines whether a particle will be climbed or not.
- The maximum particle size where there is enough time for dislocations to climb across particles is referred to as the critical radius r_{crit} .
- For smaller particles, there is sufficient time for the dislocations to climb across them. These particles will not contribute to the creep strength.
- Particle shearing is not considered since the dislocations are assumed to climb across small particles freely.
- Larger particles have to be passed by Orowan bowing, which determines their contribution to the strength.

These principles have been applied to austenitic stainless steels. It has been possible to predict the total creep strength quite accurately [14, 20, 21, 29].

For climb to be of significance, the time for a dislocation to climb across a particle t_{climb} must be at least as long as the glide time between the particles t_{glide} . This criterion can be used to find the critical radius.

$$t_{climb} = t_{glide}. \quad (10)$$

The climb time is equal to the critical radius r_{crit} divided by the climb velocity v_{climb}

$$t_{climb} = \frac{r_{crit}}{v_{climb}}. \quad (11)$$

The climb velocity is proportional to the climb mobility, Eq. (6)

$$v_{climb} = M_{cl}(T, \sigma)b\sigma. \quad (12)$$

It should be noticed that it is the full applied stress that appears in Eq. (12) and not the dislocation stress, Eq. (5). The reason is that the climb rate is used to compute the time to pass the particles, i.e. their effective resistance against climb. If the dislocation stress would be used, the influence of the particles would be taken into account twice. The necessity of this assumption is also illustrated by the fact that if the dislocation stress would be used, the resulting critical radii are so small that the particles would only have a marginal effect on the creep rate in distinct contrast to the influence on the observed rates. In previous versions of the model, the dislocation stress is applied in Eq. (12). This is now considered to be a mistake. The reason that it did not affect the results much was that the internal stress from the particles was only a smaller part of the total creep strength. For the Cu–Co system this is not the case as will be seen below. The use of the full applied stress should also be used in the climb mobility in Eq. (12) as indicated in the equation. For the data of the Cu–Co system, this has only a minor effect due to the comparatively high temperature. However, when computing the average strain rate in the material, the dislocation stress should be used, Eq. (4); otherwise, the role of the particles would not be taken into account.

The glide time is given by the interparticle spacing λ divided by the glide velocity v_{glide}

$$t_{glide} = \frac{\lambda}{v_{glide}}. \quad (13)$$

The glide velocity can be found from the Orowan equation for the creep rate

$$\dot{\epsilon} = v_{glide} \frac{b\rho}{m} \quad (14)$$

where ρ is the dislocation density. Combining Eqs. (10)–(14) gives an expression for the critical radius.

$$r_{crit} = 2M_{cl}(T, \sigma)b^2\sigma\lambda_F \frac{\rho}{\dot{\epsilon}_{sec}m}. \quad (15)$$

The secondary creep rate in Eq. (4) is used for the creep rate in Eq. (15), but with the dislocation stress, Eq. (5) is used instead of the applied stress. In Eq. (15), the Friedel spacing λ_F is introduced, which is a change in relation to previous versions of the model. It is believed to represent the actual spacing of particles along the dislocation line better than the planar square lattice particle spacing λ_s [1, 5]. The concept of Friedel spacing is fully explained in [1]. λ_F depends on the force F acting on a climbing segment

$$\left(\frac{\lambda_s}{\lambda_F}\right)^2 = \frac{F}{2\tau_L} = \frac{\sigma_{clmin} b \lambda_F}{2m\tau_L} = \frac{\sigma_{clmin}}{\sigma_O} \frac{\lambda_F}{\lambda_s}. \quad (16)$$

From Eqs. (8) and (16), we find that

$$\left(\frac{\lambda_s}{\lambda_F}\right)^3 = \frac{\alpha}{\alpha + 2C}. \quad (17)$$

With this relation, the Friedel spacing can be obtained directly.

Size distributions

The authors have found that precipitates in creep resistant steels often form exponential size distributions, see for example [21, 30]. The number of particles per unit area N_A can be described by

$$N_A = N_{A0} e^{-k(r-r_0)} \quad (18)$$

where $N_{A0} = 1/\lambda_s^2$, and r is the particle radius. r_0 is a small quantity taking into account that there is often no reliable observations at very small particle sizes; r_0 is taken as 1 nm. k is related to the average particle size $\bar{r}$: $k = 1/(\bar{r} - r_0)$. As emphasized above, only particles larger than r_{crit} are assumed to contribute to the creep strength. The average spacing between such particles is referred to as λ_{crit} .

$$\lambda_{crit} = \sqrt{N_{A0}} e^{-k(r_{crit}-r_0)/2}. \quad (19)$$

Assuming that particles larger than r_{crit} contribute to the creep strength $\sigma_{partcreep}$ through the Orowan mechanism, we find that

$$\sigma_{partcreep} = \frac{CGbm}{\lambda_{crit}} = \sigma_O e^{-k(r_{crit}-r_0)/2}. \quad (20)$$

Since r_{crit} depends on temperature and applied stress, so does $\sigma_{partcreep}$. $\sigma_{partcreep}$ is the internal stress that should be inserted in Eq. (5).

Solid solution hardening

The investigated alloys contain 0.33 wt% Co in solid solution, see Table 1. For completeness, the contribution to the creep strength will be calculated in spite of the low Co content. The linear size misfit between Co and Cu atoms is 1.28% [31]. The principles for solid solution hardening during creep are presented in [32]. The maximum interaction energy between a solute and a dislocation is $U_{max} = 2.88 \times 10^{-40}$ J. The value of β is $U_{max}/b = 1.13 \times 10^{-30}$ J/m. The precise expression for the solid solution hardening depends on the radius p

$$p = \frac{\beta}{k_B T}. \quad (21)$$

If p is less than the core radius which is taken as b , the following expression applies for the amount of solid drag [26]

$$\sigma_{drag} = \frac{v_{glide} c_0 \beta^2}{D_{Co} k_B T b} \ln\left(\frac{D_{Co}}{v_{glide} b}\right) \quad (22)$$

where v_{glide} is given by Eq. (14), c_0 is the amount of Co in solid solution (atom fraction), and D_{Co} is diffusion coefficient for Co in Cu. The resulting solid solution hardening lies between 0.15 and 0.25 MPa for the Cu–Co alloys. This low value is neglected in the analysis below.

Results

Pure copper

Except for the creep tests for Cu–Co alloys at 439 °C, creep tests for oxygen free pure copper (CuOF) were performed at the same temperature [15, 16]. In addition, creep data for CuOF have been taken from Ref. [33] at different temperatures. A comparison between these experimental data and the model in Eq. (4) is shown in Fig. 1. It is obvious that the temperature dependence of the creep rate is larger in the model than in the experiments. The reason is that the creep activation energy is smaller than that for self-diffusion in this temperature range, which has not been fully explained. However, the accuracy of the prediction around 439 °C is sufficient for a meaningful comparison to Cu–Co alloys.

Figure 1 Modelling of stationary creep rate (Eq. 4) for CuOF at different test temperatures compared with experimental data (Experimental data from [33] and [15]).

Cu–Co alloys

As discussed in the section on precipitation hardening, the maximum radius r_{crit} for which the dislocations have time to climb across the particles, play an important role. Exponential size distributions are assumed, since in previous studies, such size distributions were found in creep resistant steels both for the austenitic stainless steels 310NbN and Sanicro 25 as well as for the Cr–Mo-steels P91 and P92 [13, 21]. The assumed exponential size distributions Cu0.88Co are illustrated in Fig. 2. The total number of particles smaller than a given radius per unit area is shown (cumulative size distribution). Critical radii of

Figure 2 Size distributions (Eq. 18) for Cu0.88Co particles in aged, underaged and overaged conditions with critical radii marked (Eq. 15).

particles at 439 °C for the experimental stress ranges are calculated by Eq. (15).

The internal stress from the particles (critical Orowan stress) according to Eq. (20) is illustrated in Fig. 3. The internal stress increases with applied stress. This means that r_{crit} decreases with increasing applied stress as pointed out above. The difference between the internal stress and the applied stress and the ratio between them increase with increasing applied stress.

The critical radius decreases with increasing applied stress as shown in Fig. 4. This means that the internal stress increases with applied stress as illustrated in Fig. 3. However, the applied increases faster than the internal stress, Fig. 3, and consequently the effective stress and thereby also the creep rate increase with increasing applied stress.

In Fig. 5, the ratio between the critical Orowan stress, Eq. (20) and the ordinary Orowan stress, Eq. (7), is given as a function of the applied stress. It is illustrated that the ratio varies between 0.08 and 0.4. The ratio increases with applied stress and with the cobalt content, since the particle size is raised with cobalt content.

The creep rates for the Cu–Co alloys can now be determined. Equation (4) for the creep rate of pure copper is used but with the dislocation stress $\sigma - \sigma_i$, Eq. (5). σ_i is then given by Eq. (20). The result is illustrated in Fig. 6, where the modelled results are compared with the experimental data of Cu–Co alloys. For the considered Cu–Co alloys, the

Figure 3 Critical Orowan stress (Eq. 20) versus applied stress for Cu–Co alloys. For comparison, a 1:1 line for the applied stress is included in the diagram.

Figure 4 Critical radius (Eq. 15) versus applied stress for Cu–Co alloys.

Figure 5 Ratio between critical Orowan stress (Eq. 20) and the Orowan stress (Eq. 7) versus applied stress for Cu–Co alloys.

precipitation hardening has quite a dramatic effect. The creep rate of Cu–Co alloys is two orders of magnitude or more below that of pure copper. With increasing Co content, the creep rate decreases. This is a consequence of the denser particle distribution of the alloys with higher Co content.

For pure copper, the effective stress is equal to the applied stress, i.e. the internal stress vanishes. The lower creep rate of the Cu–Co alloys is due to the internal stress. Let us take an example and consider an applied stress of 60 MPa for the Cu2.48Co alloy. According to Fig. 3, the internal stress is 45 MPa in this case. Thus, the effective stress is only 15 MPa.

Figure 6 Modelling of stationary creep rate (Eq. 4) for three Cu–Co alloys and pure copper compared with experimental data from [15].

This illustrates why the creep rate for the Cu–Co alloys is so low.

In Fig. 6, the slope of the curves for Cu–Co alloys increases with applied stress, i.e. the creep stress exponent is raised. There are two reasons for this behaviour. First, the dislocation stress $\sigma - \sigma_i$ increases faster than the applied stress. This is evident from Fig. 3. Second, the expression for dislocation mobility in Eq. (6) increases with applied stress. This is due to the strain-induced increase in vacancy concentration at higher stresses [18, 19].

The creep data in [15] gives another possibility for comparison to the model. The Cu0.88Co alloy was investigated at different ageing times, corresponding to underaged, aged and overaged conditions, see Table 1. The comparison with the model is given in Fig. 7. Again the model can represent the observations in a reasonable way.

It is illustrated in Fig. 5 that there is no simple relation between the room temperature Orowan stress given in Table 1 and the internal stress during creep, which is referred to as the critical Orowan stress. For the underaged condition in Fig. 7, this is even more evident since the room temperature strength is controlled by particle shearing. The particle shearing strength decreases with decreasing particle radius and that is the reason why this condition has a lower hardness, as shown in Table 1. The fine particles in the underaged state give a smaller

Figure 7 Modelling of stationary creep rate (Eq. 4) for Cu0.88Co in underaged, aged and overaged conditions compared with experimental data from [15].

contribution to the creep strength than for the other alloys since the dislocation can climb the fine particles without resistance. That the Orowan stress is very large does not have any influence.

Discussion

There is an extensive literature on the strengthening mechanism of particles during creep. The simplest model is to assume that the contribution from the particles is given by the Orowan stress. The values for the Orowan stress are given in Table 1. Modelled values that at least approximately represent the experimental internal stresses can be found in Fig. 3. A direct comparison is given in Fig. 5 for the three Cu–Co alloys. The ratio between the internal stress during creep and the Orowan stress is 0.08–0.4. It is evident that the Orowan stress grossly overestimates the internal stress. In addition, the Orowan stress does not even rank the alloys in the right order.

Many attempts have been made in the literature to estimate a threshold stress for climb. The consensus now seems to be that this threshold stress is quite low and approximately given by Eq. (8). For an alloy with a particle volume fraction of 2.5%, this gives a threshold stress of $0.04 \sigma_O$, where σ_O is the Orowan stress. This is quite a low value and cannot represent all of the precipitation hardening. For the Cu2.48Co alloy, σ_O is 473 MPa (Table 1). With a particle volume fraction of 2.2%, a threshold stress of the $0.04 \sigma_O = 19$ MPa would be expected. However, at

low stresses, the computed internal stress in Fig. 3 is 25 MPa. This value is thus very close to the predicted climb threshold. But there is no indication in the curve in Fig. 3 that it is close to a threshold stress. If the low stress is close to a threshold stress, the curves would have bent down. Thus, even a threshold stress of $0.04 \sigma_O$ is too large to be consistent with the experiments. Consequently, the meaning of the climb threshold stress is unclear.

There are other reasons why a constant threshold stress is in conflict with experiments except for some oxide dispersion strengthened (ODS) alloys. If Eq. (1) is applied with a constant stress σ_i , the creep exponent would decrease with increasing applied stress and that is not observed in general. This was further discussed in the introduction. For example, in Figs. 6 and 7, the creep exponent increases with applied stress for all the curves. It is well established that for many particle strengthened steels, the creep strength and consequently the inverse creep rate using Monkman–Grant relation decrease approximately exponentially with increasing temperature and decreasing applied stress [34]. This is clearly inconsistent with a constant threshold stress.

In the present paper, the amount of climb is used to explain the precipitation hardening mechanism. In the model, the climb rate is related to the fraction of particles and their size distribution. The actual amount of climb is based on a simple assumption by comparing the time for dislocation climb and glide. One key advantage of the present model is that it can be used to explain that the creep strength decreases exponentially with increasing temperature [14, 20, 21, 29]. Also, the stress dependence of the creep rate is in agreement with observations, see Figs. 6 and 7.

Attempts have also been made for ODS alloys to explain the temperature and stress dependence. In [35], a model for creep in ODS alloys was formulated. Instead of a true threshold stress, the model introduced a parameter, a relaxation factor k , taking into account the temperature dependence of the attractive interaction between dislocation and dispersion particles. However, there were restrictions. The relaxation factor k was handled as an adjustable parameter. For estimating the k value, at least five creep tests under two temperatures and stresses were necessary. To increase the accuracy, even more tests were needed. Reppich [36] then improved the attractive particle dislocation

interaction theory. He treated the dislocation bypass as a serial process from detachment controlled to local climb controlled process. Instead of a constant threshold stress, it was calculated for detachment threshold and Orowan stress separately. The resulted overall threshold stress lied between 0.15 and 0.5 σ_O . However, the improvement was not dramatic and an adjustable k was still used.

Conclusions

A previously developed model for the particle contribution to the creep strength is critically tested by comparison to published data for Cu–Co alloys. This alloy system is particularly useful for analysing precipitation hardening, because (1) particles generate the main strengthening contribution, (2) only one type of particles is present homogeneously distributed, (3) the particles are stable during creep (after a suitable heat treatment), and (4) the amount of elements in solid solution is limited.

Precipitation hardening is in general the most potent way of increasing the creep strength of alloys. Consequently, an extensive literature on the topic is available. In spite of this, few predictive models are available that can describe the observations. For this reason, an attempt was made to critically test the model by Eliasson et al. Two changes in the model are made. The climb speed when particles are passed is now assumed to be controlled by the full applied stress, not just the dislocation stress. The Friedel particle spacing is used instead of the planar lattice square spacing when computing the critical particle radius. The model is based on the following assumptions

- The controlling mechanism is the time it takes for a dislocation to climb across a particle.
- The key quantity of the model is the critical particle radius. Above this radius, the particles cannot be passed by climb.
- An exponential distribution of particle sizes is assumed following results for creep resistant steels. For small particles, since enough time exists for dislocations to climb, they do not contribute to the creep strength. Only particles big enough that cannot be climbed during the creep life contribute to the strength. This strength

contribution is calculated with the Orowan mechanism.

- The model is fully predictive in the sense that no adjustable parameters are used.

The study gave the following results

- In some investigations, the Orowan stress is used to estimate the creep strength. It is demonstrated that this grossly overestimates the contribution to the creep strength for the Cu–Co alloys.
- In the past, quite a low value of about 0.03 σ_O for a climb threshold stress has been derived, where σ_O is the Orowan stress. However, even this low value is too large to be consistent with the data for one of the Cu–Co alloys.
- The model can describe the influence of applied stress, alloy composition and heat treatment for the Cu–Co alloys at least approximately. Previous studies for austenitic stainless steels demonstrate that the model can represent the temperature dependence.
- The model gives a stress and temperature dependent internal stress from the particles, which is a requirement to describe the experiments (see the previous bullet). For example, a constant internal stress would give a stress exponent that decreases with increasing applied stress, which is inconsistent with observations except for some ODS alloys.

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References

- [1] Martin JW (2012) Precipitation hardening: theory and applications. Butterworth-Heinemann, Oxford
- [2] Lagneborg R (1973) Bypassing of dislocations past particles by a climb mechanism. *Scr Metall* 7:605–613
- [3] Lagneborg R, Bergman B (1976) The stress/creep rate behaviour of precipitation-hardened alloys. *Met Sci* 10:20–28
- [4] Arzt E, Ashby MF (1982) Threshold stresses in materials containing dispersed particles. *Scr Metall* 16:1285–1290
- [5] Blum W, Reppich B (1985) Creep of particle-strengthened alloys. In: Wilshire B, Evans RW (eds) *Creep behaviour of crystalline solids*. Pineridge Press Ltd., Swansea, pp 83–135
- [6] Heilmaier M, Reppich B (1999) Particle threshold stresses in high temperatures yielding and creep: A critical review. In: *The minerals, metals and materials society: creep behavior of advanced materials for the 21st century*, pp 267–281
- [7] Rösler J, Arzt E (1988) The kinetics of dislocation climb over hard particles. I. Climb without attractive particle-dislocation interaction. *Acta Metall* 36:1043–1051
- [8] Kelly A, Nicholson RB (1971) Strengthening methods in crystals. Materials science series. Applied Science Publishers Ltd., London
- [9] Shewfelt R, Brown L (1977) High-temperature strength of dispersion-hardened single crystals II: theory. *Philos Mag* 35:945–962
- [10] Nardone VC, Tien JK (1983) Pinning of dislocations on the departure side of strengthening dispersoids. *Scr Metall* 17:467–470
- [11] Schröder JH, Arzt E (1985) Weak beam studies of dislocation/dispersoid interaction in an ods superalloy. *Scr Metall* 19:1129–1134
- [12] Eliasson J, Gustafson Å, Sandström R (2000) Kinetic modelling of the influence of particles on creep strength. *Key Eng Mater* 171–174:277–284. <https://doi.org/10.4028/www.scientific.net/KEM.171-174.277>
- [13] Magnusson H, Sandström R (2007) The role of dislocation climb across particles at creep conditions in 9 to 12 pct Cr steels. *Metall Mater Trans A* 38:2428–2434
- [14] Vujic S, Sandstrom R, Sommitsch C (2015) Precipitation evolution and creep strength modelling of 25Cr20NiNbN austenitic steel. *Mater High Temp* 32:607–618
- [15] Threadgill P, Wilshire B (1974) The effect of particle size and spacing on creep of two-phase copper–cobalt alloys. *Met Sci* 8:117–124
- [16] Parker J, Wilshire B (1975) The effect of a dispersion of cobalt particles on high-temperature creep of copper. *Met Sci* 9:248–252
- [17] Sandstrom R, Andersson HCM (2008) Creep in phosphorus alloyed copper during power-law breakdown. *J Nucl Mater* 372:76–88
- [18] Spigarelli S, Sandström R (2018) Basic creep modelling of aluminium. *Mater Sci Eng A* 711:343–349
- [19] Sandström R (2017) Fundamental modelling of creep properties. In: Tomasz Tanski MS, Zieliński A (eds) *Creep*. InTech, Rijeka
- [20] Sandström R (2016) Fundamental models for creep properties of steels and copper. *Trans Indian Inst Met* 69:197–202
- [21] Sandström R, Farooq M, Zurek J (2013) Basic creep models for 25Cr20NiNbN austenitic stainless steels. *Mater Res Innov* 17:355–359
- [22] Sandström R, Hallgren J (2012) The role of creep in stress strain curves for copper. *J Nucl Mater* 422:51–57
- [23] Sandstrom R (2012) Basic model for primary and secondary creep in copper. *Acta Mater* 60:314–322
- [24] Sui FF, Sandstrom R (2018) Basic modelling of tertiary creep of copper. *J Mater Sci* 53:6850–6863. <https://doi.org/10.1007/s10853-017-1968-7>
- [25] Sandström R (2016) The role of cell structure during creep of cold worked copper. *Mater Sci Eng A* 674:318–327
- [26] Hirth JP, Lothe J (1982) *Theory of dislocations*. Krieger, Malabar
- [27] Kocks U (1966) A statistical theory of flow stress and work-hardening. *Philos Mag* 13:541–566
- [28] Shewfelt RSW, Brown LM (1977) High-temperature strength of dispersion-hardened single crystals II: theory. *Philos Mag* 35:945–962
- [29] He J, Sandström R (2017) Basic modelling of creep rupture in austenitic stainless steels. *Theor Appl Fract Mech* 89:139–146
- [30] Magnusson H, Sandstrom R (2007) The role of dislocation climb across particles at creep conditions in 9 to 12 pct Cr steels. *Metall Mater Trans Phys Metall Mater Sci* 38A:2428–2434
- [31] King HW (1966) Quantitative size-factors for metallic solid solutions. *J Mater Sci* 1:79–90. <https://doi.org/10.1007/BF00549722>
- [32] Sandström R, Korzhavyi PA (2014) Modelling the contribution from solid solution hardening to the creep strength of austenitic stainless steels. In: *10th Liège conference on materials for advanced power engineering*, Liège
- [33] Raj S, Langdon T (1989) Creep behavior of copper at intermediate temperatures. I. Mechanical characteristics. *Acta Metall* 37:843–852
- [34] Abe F, Kern T-U, Viswanathan RE (2008) *Creep-resistant steels*. Elsevier, New York

- [35] Rösler J, Arzt E (1990) A new model-based creep equation for dispersion strengthened materials. *Acta Metall Mater* 38:671–683
- [36] Reppich B (1998) On the attractive particle–dislocation interaction in dispersion-strengthened material. *Acta Mater* 46:61–67