PARTICLE THRESHOLD STRESSES IN HIGH TEMPERATURE YIELDING AND CREEP: A CRITICAL REVIEW

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Abstract

At low temperatures incoherent non-shearable oxide particles introduce a yield strength increment for dislocation glide due to Orowan bypassing, called threshold stress $\sigma_{th}$. However, at high temperatures such a true threshold does not exist. In contrast, the material deforms even under the lowest stress applied. Therefore, the Orowan process has lost its predominating microstructural significance. Instead, models for climb threshold developed in the past with respect to different particle shapes and climb geometries will be discussed by introducing a universal parameter called climb resistance $R$. An essential result for most oxide dispersion strengthened (ODS) alloys is that interfacial pinning, i.e. the detachment of the partially relaxed dislocation from the attractive particle-matrix interface controls the creep kinetics. As a consequence, our concept tackles creep thresholds in three steps: first, we associate $\sigma_{th}$ with the "apparent" particle hardening contribution $\sigma_p$ to realize a creep rate $\dot{\varepsilon}$ in the particle-strengthened alloy equivalent to that of the corresponding single-phase matrix. Second, the course of the "true" particle threshold stress $\sigma_0^*$ with respect to $\dot{\varepsilon}$ is modelled in terms of the operating elementary particle hardening mechanisms. Finally, $\sigma_0^*$ is introduced into the kinetic creep law. We exemplify and verify our approach with selected ODS platinum-, nickel- and iron-base superalloys.
Introduction

In the past 40 years there has been widespread and growing consensus that plastic deformation of materials at high temperatures is not driven by the full applied stress but rather by a reduced stress. This effective stress reflects the resistance of the material to deformation and is invariably calculated by subtracting an internal back stress or threshold stress from the applied stress:

\[ \sigma_{\text{eff}} = \sigma - \sigma_t = \sigma - \sigma_{\text{th}}. \]  

The quantity to be subtracted in this way may depend on the composition, the microstructure, the thermomechanical history, the temperature, the creep strain and the imposed stress or strain rate. In the deformation of solids this approach has been applied to the study of diffusional, superplastic, dislocation and dislocation-particle interaction processes. It has become particularly popular in the development of constitutive equations used to describe material behaviour under diverse conditions.

Unfortunately the apparent success of the effective or threshold stress concept may conflict with attempts in deriving a fundamental understanding of the underlying deformation mechanisms. Thus, popularity has resulted in a somewhat indiscriminate use of this approach with attendant confusion.

The present paper is an attempt to fill some gaps left by a series of excellent overviews [1, 2, 3, 4, 5, 6] with particular emphasis on dispersion strengthening in ODS alloys arising from incoherent, non-shearable hard ceramic particles. Consequently, we will replace the general term \( \sigma_{\text{th}} \) by the more specific expression \( \sigma_p \), (where the index "p" is the abbreviated form of "particles") in the following.

The Particle Threshold Stress Concept

At low temperatures oxide dispersoids introduce a yield limit or threshold stress \( \sigma_p \) due to Orowan by-passing below which no dislocation glide is possible. This concept of a "true" threshold stress has been first extended to elevated temperatures by Brown and Ham [2]. Later, Lund and Nix [7] gave a first microstructural interpretation of the experimentally determined threshold stress on thoria dispersed (TD) Nichrome in terms of the Orowan stress \( \sigma_{\text{OR}} \) by the following linear superposition rule with the creep strength of the particle-free matrix \( \sigma_m \):

\[ \left( \frac{\sigma}{G} \right)_{LT} = \left( \frac{\sigma_m}{G} \right)_{LT} + \left( \frac{\sigma_{\text{OR}}}{G} \right)_{LT}, \]  

for any given constant strain rate \( \dot{\varepsilon} \) and temperature \( T \). However, at high temperatures such a "true" threshold stress does not exist. In contrast, the material deforms even at the lowest stress applied [8, 9]. This situation is exemplified in fig. 1 revealing the following main results: (i) at low temperatures the yield stress increment \( \Delta R_{0.2} \) at 0.2% plastic strain due to oxide dispersoids in MA 754 (taking Nimonic 75 as an appropriate representative for the particle-free matrix) coincides with the anisotropic Orowan stress acc. to Kcock's formula [11]:

\[ \sigma_{\text{OR}} = 0.9M[\ln(2\pi t/b)]^{1/2} \frac{K^e}{b(L - \frac{3}{2}r)}, \]  

with the prelogarithmic line tension factor of a straight edge dislocation \( K^e = \frac{G}{2\pi}(r+D) \), the Burgers vector \( b \), and the Taylor factor (or reciprocal Schmid factor) \( M \). \( r \) and \( L \) are the mean radius and mean planar centre-to-centre spacing of the particles, respectively. (ii) At high temperatures, specifically above 800°C, \( \Delta R_{0.2} \) falls below the shaded band for the Orowan process to yield roughly only 1/3 to 1/2 of \( \sigma_p \) indicating a change from glide- to diffusional climb-controlled overcoming of the oxide dispersoids [2, 4].

![Figure 1: Comparison of creep strength and dislocation configuration of ODS MA 754 between room temperature and 1200°C: (a) compressive 0.2% yield stress of MA 754 (full circles) and Nimonic 75 (open circles). Shaded: Orowan stress calculated acc. to eq. 3 as low temperature yield stress increment due to oxide dispersoids [6]. (b) Percentage of dislocations looped and attached to particle, respectively [10].](image)

The Climb Threshold

It is widely accepted that at high temperatures dislocations undergo non-planar motion by climb allowing for a dislocation segment arrested at a particle to bulge out of the slip plane and surpass the particle. The origin of the threshold stress is then the increase in dislocation line length during by-pass. Fig. 2 summarizes the various current models on climb thresholds \( \sigma_{\text{pc}} \) with respect to different climb geometries and particle shapes. Using the Peach-Koehler approach Blum and Reppich [4] demonstrated that the climb thresholds \( \sigma_{\text{pc}} \) in all models could be uniquely described by introducing an unified parameter \( R_c \) called climb resistance [18], see Fig. 2,

\[ R_c = \left( \frac{dl}{dy} \right)_{\text{max}} = \frac{2\sigma_{\text{pc}}}{(MGb/L)} \]  

which represents the maximum value of line length (or line energy, alternatively) increase \( dl/dy \) as the dislocation proceeds over the particle in y-direction, \( G \) is the shear modulus.
Local climb

(ii) General climb

(a) Brown and Ham [2]; (b) and (e) Shewfelt and Brown [12] and Stevens and Flewitt [13]; (c) Lagneborg [14]; (d) and (e) Hausseit and Nix [15]; (f) Evans and Knowles [16]; (i) local and (ii) general climb [18]. Note that, as pointed out by Mishra and Mukherjee [17], the "zig-zag"-shaped configuration is not justified in the case of a climbing edge dislocation. Note, that the term \( \text{MGb} / \text{L} \) used for normalization in eq. 4 represents the classical Orowan stress which will be shown to be common for all threshold stress models discussed subsequently. It is obvious from Fig. 2 that all models can be grouped into two classes: during local climb the dislocation remains in the particle/matrix interface, Fig. 2i whereas in general climb models the dislocations are allowed to unravel from the interface, and hence, to further reduce their line length, Fig. 2ii. Interestingly, the former models commonly predict a constant threshold dependent on the shape but independent on the volume fraction of the particles under consideration, see Figs. 2a and b. In contrast, all latter models on general climb reveal a dependence on volume fraction.

Interfacial Pinning

This mechanism has become particularly popular in the past 10 years or so as it seems to have the capacity for explaining the unusual \( \dot{\varepsilon} - \sigma \) creep characteristics of ODS alloys, e.g. high stress exponents \( n \) and activation energies \( Q_{\text{amp}} \). It is essentially based on the assumption of an attractive interaction between the gliding/climbing matrix dislocation and the incoherent interface of the particle [19]. Hence, particle overcoming has to be treated as a serial process of the dislocation consisting of local climb over (which has been predicted to be stabilized by the attraction [20]) and detachment from the departure side of the particle, Fig. 3a. Indeed, dislocations pinned to the particle's interface as displayed in

![Figure 3](image-url)

Figure 3: The mechanism of interfacial pinning (IP): (a) perspective view illustrating the serial process of local climb over and detachment from the particles; (b) TEM micrograph showing evidence of IP in crept PM 2000; (c) Top view revealing the force balance and the dislocation configuration in the moment of detachment, see text [26].

Fig. 3b have been observed frequently in the literature to justify a particular importance of this mechanism, see for example [10, 21, 22]. Following the early suggestion by Blum and Reppich [4] that the physical origin of the attractive interaction is based on the line energy reduction of the dislocation segment residing in the particle/matrix interface, Rosler and Arzt developed a kinetic equation for detachment-controlled creep (in the following "RA-model") [23]:

\[
\sigma_{\text{a}}(\dot{\varepsilon}, T) = \frac{\dot{\varepsilon}}{\frac{k_B T}{\ln \dot{\varepsilon}}} \left( \frac{1}{1 - k} \right) \cdot \left( \frac{1}{1 - k} \right)
\]

(5)

The reference strain rate is given as \( \dot{\varepsilon}_0 = \frac{3D_{\text{V}} \rho}{k_B N} \), where \( D_V \) is the volume diffusion coefficient and \( \rho \) is the dislocation density. The "athermal" detachment stress \( \sigma_0 \) reads

\[
\sigma_0 = \sqrt{1 - k^2} \cdot (\text{MGb}/\text{L})
\]

(6)

Unfortunately, in this approach the relaxation factor \( k = E_0 / E_{\text{app}} \) which is defined as the ratio between the dislocation line energies at the interface and within the bulk matrix, is used as a free adjustable parameter only and not modelled from first principles. A rather weak reduction of the line energy in the particle/matrix interface of 6% yielding \( k = 0.94 \) favors the detachment process to offset local climb and, thus, to be rate controlling.

The creep behavior of dispersion strengthened alloys has been indeed found to be consistent with the features of eq. 5 with respect to the general interrelations between \( \sigma, \dot{\varepsilon}, T \) and the
A promising way out of the dilemma with the somewhat arbitrarily use of the relaxation factor has been shown by Mishra et al. [27] considering the dissociation of the matrix dislocation into partial interface dislocations. Then, the reduction of dislocation self-energy resulting from dissociation gives $E_{\text{dis}}/E_m \leq 0.64 (= k)$, a rather high value which is in obvious contradiction to the result of the RA-model [24]. In addition, their model also predicts a threshold stress whose magnitude depends on particle radius and interparticle spacing. However, its distribution so far has been limited due to the need of numerically solving the equation for the threshold stress and due to the lack of TEM observations supporting the assumption of dislocation dissociation. Arzt and Gohring [28] extended the detachment model, eqs. 5 and 6 for the case of a superdislocation in an ordered matrix taking into account additional interactions resulting from repulsive forces between the two superpartials and from attraction due to the presence of antiphase boundaries. Finally, Dunand and Jansen [29] presented a new model which extends the former models to high volume fractions of particles by taking into account the effect of dislocation pile-ups on the detachment process. However, as the present paper focuses on disordered ODS alloys with a low volume fraction of dispersoids $\leq 3\%$, the latter models may not be discussed further here.

**Experimental Determination of the Threshold Stress**

Modeling the creep behavior of real alloy systems requires not only the development of adequate microstructural models, as elucidated above, but also the exact determination of the threshold stresses $\sigma_p$. The following methods are well-established in the literature:

**Parameter Fit**

It starts from the classical threshold stress equation [4]

$$\dot{\epsilon} = A \left(\sigma - \sigma_p\right)^n$$  \hspace{1cm} (9)

considering the obstacles to dislocation movement as serial elements consisting of the hardening contribution $\sigma_p$ from the overcoming of the particles and of the matrix glide resistance ($\sigma_m = \sigma - \sigma_p$). Consequently, the material is assumed to deform in response to the reduced stress ($\sigma - \sigma_p$) and e.g. $n_m = 4$ might be taken as an appropriate value for the constant stress exponent of the matrix (note the obvious analogy to eqs. 1 and 2). Furthermore, $\sigma_p$ is considered as a free, adjustable parameter only resulting from the respective fit of $\dot{\epsilon}$ vs. ($\sigma - \sigma_p$). As an example, fig. 5a compares selected ODS nickel-base alloys (e.g. the "MA" family) with "conventional" $\gamma'$-strengthened nickel-base alloys (e.g. Udimet 700). Obviously, at the chosen temperature of 760°C the coherent, ordered $\gamma'$ particles (present in high volume fraction of roughly up to 70% in the case of single crystals [30]) are superior strengtheners. However, this behavior can be reversed at temperatures in excess of 1000°C where the oxide dispersoids remain thermally stable [31]. In contrast, the $\gamma'$ particles suffer from drastic coarsening and rafting [30] as well as from a significant reduction of their volume fraction as creep temperatures approach the $\gamma'$-solvus.

**Lagneborg-Bergman Plot**

A more generalised approach can be obtained by rearranging eq. 9 acc. to Lagneborg and Bergman [32]

$$\dot{\epsilon}^{1/n_m} = A^{1/n_m} \left(\sigma - \sigma_p\right)$$  \hspace{1cm} (10)
\[ \frac{n}{k} = \text{const.} \]

\[
\sigma_p = \alpha G b M \sqrt{\rho}
\]

where \( \sigma_p \) is the stress component of single dislocations defined as the Taylor back stress with the elastic interaction constant \( \alpha \) ranging from 0.1 \ldots 1. The back stress \( \sigma_p^{\text{back}} \) originates from the evolution of a heterogeneous dislocation structure consisting of hard subgrains and soft subgrain interior as successfully applied in the composite model of plastic deformation [37]. The threshold stress \( \sigma_p \) should now be associated with the "apparent" particle hardening contribution to realize a creep rate \( \dot{\varepsilon} \) in the particle-strengthened alloy equivalent to that of the corresponding single-phase matrix. Note, that not all contributions to the internal stress \( \sigma_i \) appearing in eq. 11 have to be present simultaneously: for example in particle-free matrix materials obviously \( \sigma_p = 0 \) whereas in ODS alloys the evolution of subgrains is often retarded, thus \( \sigma_p^{\text{back}} = 0 \) in this case [36]. An experimental example of this generalized approach on the determination of \( \sigma_p \) and subsequent description of the steady-state creep behavior is shown in Fig. 6 with ODS platinum-base alloys1. Modeling acc. to eq. 11 involves several steps: first, the creep behavior of pure Pt has been

\[ \varepsilon_{\text{HAI}} = \frac{\rho b v_0}{M} \sinh(\beta (\sigma - \sigma_p - \sigma_p^{\text{back}} - \sigma_p)). \]  

\[ \sigma_p = \alpha G b M \sqrt{\rho} \]

Figure 5: (a) Double-logarithmic plot of creep rate vs. reduced stress \((\sigma - \sigma_p)\) for selected nickel-base superalloys. (b) Schematic Lagneborg-Bergman plot [4, 34].

Figure 6: Comparison of the creep behavior of ZrO\(_2\) dispersion strengthened Pt-base alloys at 1250°C: (a) Double-logarithmic Norton plot of creep rate \( \dot{\varepsilon} \) vs. stress \( \sigma \); (b) Lagneborg-Bergman plot; (c) dependence of \( \sigma_p \) on \( \dot{\varepsilon} \) [6, 38]. The shaded bands denote the ratio \( \sigma_p/\sigma_{\text{on}} \).

\[ \text{The material was supplied by Degussa AG, Hanau, Germany. The Zirconia dispersoids have been introduced immediately after an immediate change of stress provided that it has been deformed in a steady-state condition before [35].} \]
described, see the left curves in Fig. 6a (Norton plot) and b (Lagneborg-Bergman plot), respectively. Note, that in contrast to the simplified approach sketched in Fig. 5b the matrix curve indeed reveals curvature indicating a transition to the power law breakdown regime. Second, the particle hardening contribution \( \sigma_p \) is determined at any given creep rate simply as the horizontal stress difference in Fig. 6b. It is obvious that only a few data points at high strain rates could be approximated by a straight line independently on the value of \( n_p \) chosen. Instead, the majority of the experimental data points deviates to lower stresses when approaching the abscissa. This observation clearly indicates \( \sigma \)-proportional threshold stresses acc. to curve 2 in Fig. 5b. In turn, the dependence of the threshold stress on creep rate in Fig. 6c reveals a continuous decrease of \( \sigma_p \) with decreasing \( \dot{\varepsilon} \) and, clearly, no true threshold, where plastic deformation becomes negligible, is visible. On the other hand, at high creep rates the data points seem to level off at about 0.5 of the Orowan stress calculated acc. to eq. 3. With the appropriate \( \sigma_p \)-values obtained from the Lagneborg-Bergman plot, Fig. 6b, the creep curves for ODS Pt can be calculated acc. to eq. 11 yielding the continuous curves in Fig. 6a. They agree well with the measured data points. However, modeling of experimentally determined threshold stresses \( \sigma_p \) after eqs. 5 to 8 involves a refined analysis of the influence of the particles on the strain hardening behavior of ODS alloys. This modification will be presented next.

**Yield Stress Increment Method**

The classical threshold stress concept [2] assumes that no significant influence of the particles on the rate of work hardening exists. However, in contrast to single-phase material the following microstructural evidences clearly indicate that this simplified approach has to be modified: (a) the \( \dot{\varepsilon} \)-decrease during primary creep in ODS alloys is extremely high, covering usually several orders of magnitude [39, 40] (b) the steady-state dislocation density \( \rho_{ss} \) is increased at equivalent strain rates \( \dot{\varepsilon} \) [39, 40]; and (c) in constant strain rate tests the yield stress increment at zero plastic strain \( \Delta R_{p0.2} \) is significantly lower than the stress increment at steady-state \( \sigma_p \) [36], see Fig. 7. These observations indicate that the presence of the dispersoids in the matrix causes additional work hardening during plastic deformation towards steady-state. Thus, we suggest

\[
\sigma_p = \Delta R_{p0.2} - \Delta \sigma_p (\rho_{ss})
\]  

(13)

as a more appropriate measure for the "true" threshold stress \( \sigma_p \) reflecting the mere particle-overcoming process. (For practical reasons considering the limit of experimental resolution we have substituted the yield stress increment at zero plastic strain \( \Delta \sigma_p \) by its counterpart at \( \dot{\varepsilon}_t = 0.02 \)%) The particular correction term in eq. 13 accounts for the differences in the initial dislocation densities of the ODS alloy and its corresponding matrix reference alloy. It can be calculated acc. to eq. 12. Hence, \( \sigma_p \) now replaces the original threshold stress \( \sigma_{ODS} \) in eq. 11. However, the estimation of \( \sigma_p \) from measured yield stress data (Fig. 7) acc. to eq. 13 arises the question on the law of superposition of such individual hardening process affected by the interaction of the dislocations with a variety of different types of obstacles: for instance in superalloys, dislocations and solute atoms are representatives of rather weak "strengtheners" whereas particles are considerably stronger obstacles. The required yield stress increments \( \Delta R_{p0.2} \) follow from a generalized phenomenological superposition rule [6]

\[
(\Delta R_{p0.2})^m = (\tau_{ODS})^m - (\rho_{matrix})^m; \quad m = 1 \ldots 2.
\]  

(14)

Table 1: Material constants and modeling parameters for PM 2000 and MA 754 at 900 and 850°C, respectively.

<table>
<thead>
<tr>
<th>Material</th>
<th>( \alpha ) ( / \mathrm{nm} )</th>
<th>( G_s / \mathrm{GPa} )</th>
<th>( \nu_s )</th>
<th>( \rho_0 / \mathrm{m}^{-2} )</th>
<th>( r / \mathrm{nm} )</th>
<th>( L_p / \mathrm{nm} )</th>
<th>( f_p )</th>
</tr>
</thead>
<tbody>
<tr>
<td>PM 2000</td>
<td>0.16</td>
<td>0.249</td>
<td>2.65</td>
<td>59</td>
<td>0.386</td>
<td>1.0 ( \times 10^3 )</td>
<td>8.1</td>
</tr>
<tr>
<td>MA 754</td>
<td>0.23</td>
<td>0.254</td>
<td>2.45</td>
<td>56</td>
<td>0.437</td>
<td>2 ( \times 10^2 )</td>
<td>6.8</td>
</tr>
</tbody>
</table>

**Modeling of Creep Thresholds**

Figures 8a and b compare experimentally determined \( \sigma_p \)-values (symbols, eq. 13) with theoretical predictions after the RA-model (lines, eqs. 5 to 8) for the ODS alloys MA 754\(^2\) and PM 2000\(^3\). The extremal cases for either linear or pythagorean superposition acc. to eq. 14 are depicted by full and open symbols, respectively. The athermal detachment stress \( \sigma_d \) appearing in eq. 5 was calculated either after the original formulation by Röser and Arzt [23], eq. 6, or the revised version after Reppich, eq. 8 [26]. In contrast to the original RA-approach, the elastic anisotropy for cubic alloys after Chou and Sha [41] was taken into account yielding \( G_{ss} \) and \( \nu_s \) as listed in Tab. 1. Additionally, the particle and dislocation parameter as determined by TEM are summarized there.

It is obvious from Fig. 8 that generally the revised formula after Reppich, eq. 8, yields substantially lower \( k \)-values. For both materials the correlation between the experimental values (open symbols) and model curves for pythagorean superposition is poor, particularly

\(^2\)MA 754 is a trademark of Inco Alloys International, Huntington, USA. It is essentially an austenitic Yttria dispersion strengthened Nickel-Chromium solid solution made by mechanical alloying techniques.

\(^3\)PM 2000 is a trademark of Plansee AG, Reutte, Austria. It is a mechanically alloyed ferrite Yttria dispersion strengthened Iron-Chromium-Aluminum solid solution.
Figure 8: Comparison of experimental $\sigma_\tau$-values determined from yield stress data (full symbols: linear superposition, open symbols: pythagorean superposition) with theoretical predictions (lines) for selected ODS alloys: (a) nickel-base MA 754 at 850°C; (b) iron-base PM 2000 at 900°C [42].

For the Reppich approach. Moreover, the relatively steep increase of the open symbols is hardly reflected by both model variants. In contrast, linear superposition represents a much better correlation between experiment (full symbols) and calculations. While in PM 2000, Fig. 8b, the original RA approach acc. to eq. 6 seems to fit the data substantially better, no clear decision can be made in the case of MA 754, Fig. 8a: both versions describe the experimental data satisfactorily. For the matter of consistency, we will use the linear version of the original RA approach for the description of the steady-state creep behavior in the final Fig. 9.

In a final step the steady-state creep behavior of MA 754 and PM 2000 is described in Figs. 9a and b, respectively, in terms of the effective stress model, eq. 11. As pointed out above, due to the negligible amount of subgrain formation during plastic deformation the back stress resulting from subgrain hardening $\sigma_{b}^{s}$ has been set to zero for both ODS alloys. The (linear) superposition of the stress components $\sigma_{p}$, $\sigma_{\tau}$, and $\sigma_{d}$ yields the total creep strength $\sigma$ of the respective ODS alloys and agrees well with the measured data points. Surprisingly, both materials reveal that the "true" threshold stress $\sigma_{b}$ is only of the same order as the athermal back stress $\sigma_{p}$ from dislocations. Besides, the effective stress $\sigma_{e}$ which covers the solid solution friction contributes substantially to the total creep strength – particularly at high creep rates and in the case of MA 754 – and might therefore no longer be neglected in the application of detachment models [23].

Outlook

The effective stress model has been proven useful to determine the specific role of oxide dispersoids in the microstructurally based description of the creep behavior of ODS alloys. Particularly, the careful estimation of the "true" particle threshold stress $\sigma_{b}$ has allowed the identification of the RA dislocation detachment mechanism to be rate controlling during creep. A further extension of the model to transient deformation has been proven successful by incorporating an appropriate law of structural evolution for the microstructural quantities under consideration. As an example we have used a common Johnson–Mehl–Avrami kinetics for the dislocation density $\rho$ and the volume fraction of subgrains to model transient creep under constant stress as well as constant strain rate conditions of ODS nickel-base alloys like MA 754 and MA 6000 [36, 39, 40]. However, further careful experiments including appropriate creep data for particle-free matrix materials are necessary to clarify the origin and magnitude of threshold stresses in
ODS alloys. Generally, excellent model fits alone, without convincing microscopic evidence, are of limited use and do not provide a mechanistic or physically based justification for the deformation behavior of a material under consideration.

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