An Analysis of Flow Mechanisms in High Temperature Creep and Superplasticity

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The flow of crystalline solids at elevated temperatures is dependent upon creep mechanisms associated with the movement of dislocations, the relative displacements of adjacent grains and the stress-directed flow of vacancies. It is convenient to express the steady-state creep rate in terms of the dependences of these various creep mechanisms on the applied stress, the testing temperature and the grain size of the material. In practice, however, there are similarities in some of the predicted dependences for different creep processes and this may lead to experimental difficulties in unambiguously identifying the rate-controlling creep mechanism. The difficulties in identifying the creep mechanism become especially significant at low stresses. This paper reviews the characteristics of the various flow processes occurring in simple metallic systems such as pure metals and metallic alloys and describes procedures that may be adopted to provide an unambiguous identification of the rate-controlling flow mechanism.

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1. Introduction

Creep flow refers to the plastic deformation occurring over a period of time when a crystalline solid is subjected to a load which is not sufficient to cause immediate fracture. Creep is a diffusion-controlled process and therefore it becomes of major importance at elevated temperatures, typically at diffusion-controlled process and therefore it becomes of major importance at temperatures at and above \( \sim 0.5 T_m \) where \( T_m \) is the absolute melting temperature of the material.

In practice, the rate of creep depends primarily upon the testing temperature through the diffusion coefficient and upon the magnitude of the applied stress. The microstructural features of the material are also important, especially the grain size since the creep rate often increases when the grain size is reduced. The various flow mechanisms occurring during creep are usually defined in terms of the rate of flow under steady-state conditions where the strain increases uniformly as a function of time. For these conditions, it is convenient to express the steady-state creep rate, \( \dot{\varepsilon} \), by a relationship of the form:

\[
\dot{\varepsilon} = \frac{ADGb}{kT} \left( \frac{b}{d} \right)^p \left( \frac{\sigma}{G} \right)^n
\]  

where \( D \) is the diffusion coefficient (= \( D_o \exp(-Q/RT) \) where \( D_o \) is a frequency factor, \( Q \) is the appropriate activation energy for the diffusive process and \( R \) is the gas constant), \( G \) is the value of the shear modulus at the testing temperature, \( b \) is the Burgers vector of the material, \( k \) is Boltzmann’s constant, \( p \) and \( n \) are constants defined as the inverse grain size exponent and the stress exponent, respectively, and \( A \) is a dimensionless constant.

An important requirement in all creep testing is to determine the rate-controlling creep mechanism under any selected testing conditions. This evaluation requires an experimental determination of the values of \( Q \), \( p \) and \( n \) and a comparison between these values and the various theoretical models available in the creep literature. In practice, however, this comparison may be difficult, if not impossible, if several different theoretical models predict reasonably similar values. This paper addresses this problem by examining the possibility of seeking additional experimental information that may help in achieving an unambiguous identification of the rate-controlling process.

To place this paper in perspective, the following section provides a brief review of the potential mechanisms occurring in high temperature creep, sections 3 and 4 examine the mechanisms occurring at high and low stresses, respectively, and section 5 describes the additional information that is needed to achieve unambiguous identifications. For simplicity, attention is restricted to simple metallic systems such as pure metals and solid solution alloys. More complex systems, such as composite materials, may experience additional creep mechanisms related to their microstructural characteristics but these mechanisms are beyond the scope of this brief report.

2. The Flow Mechanisms Occurring in High Temperature Creep

The flow of a crystalline material is dependent upon the presence of defects in the material, where these defects may be localized at points (vacancies), lines (dislocations) or planes (grain boundaries). In broad terms, creep flow is due to the movement of dislocations within the crystalline matrix, to the relative displacements of adjacent grains in the process of grain boundary sliding and to the flow of vacancies under the action of the applied stress. Each of these flow processes is considered in turn in the following sections and each flow mechanism leads to specific predictions for the values of \( Q \), \( p \) and \( n \).

It follows from eq. (1) that the steady-state creep rate varies with the applied stress raised to a power of \( n \). Therefore, a standard procedure in creep experiments is to plot, in a double-logarithmic form, the steady-state strain rate against the applied stress. An example of this approach is shown schematically in Fig. 1. Over a wide range of high and intermediate stresses, the creep rate varies with stress raised to a reasonably high power of \( \sim 3–5 \) and in this region the flow is dominated by the intragranular movement of dislocations: this behavior is generally termed power-law
creep. At even higher stresses, the creep rate varies exponentially with stress where this behavior marks a transition from diffusion-controlled flow at the lower stresses to a thermally-activated region analogous to flow at low temperatures. This transition is generally termed the power-law breakdown and it is marked as PLB in Fig. 1. There are several possible creep mechanisms occurring at low stresses, including the stress-directed flow of vacancies in diffusion creep, grain boundary sliding (GBS) and a dislocation mechanism known as Harper–Dorn creep. As indicated by the question mark in Fig. 1, these various mechanisms tend to have fairly similar dependences on stress, temperature and grain size and accordingly it is often not easy to make an unambiguous identification of the precise rate-controlling mechanism in these low stresses.

The following section describes the behavior in the intermediate and high stress region where dislocations control the creep process and the following section describes the characteristics of flow in the low stress region where the behavior becomes more ambiguous.

3. Rate-Controlling Creep Mechanisms at Intermediate and High Stresses

The intragranular flow of dislocations controls the creep behavior at intermediate and high stresses in the region of power-law creep. An example is shown in Fig. 2 where dislocation loops are emitted and expand outwards from an array of dislocation sources, such as Frank–Read sources, located on different slip planes. Since there is an interaction between the stress fields of the leading dislocations on different but parallel slip planes, the leading screw dislocations on these different planes will have opposite signs and they readily cross-slip and annihilate. By contrast, the leading edge dislocations are blocked by these mutual interactions so that the edge dislocations pile up on each separate slip plane. Ultimately, the back stresses exerted on the sources by these pile-ups are relieved when the leading edge dislocations climb together and annihilate each other. It follows, therefore, that the annihilation process permits the emission of an additional loop from the source and in this way a regenerative process is created in which the rate of plastic flow is determined by the rate of climb. It is apparent also that this process is capable of producing very large creep strains.

An early model of creep controlled by dislocation climb incorporated the assumption that the dislocations can efficiently absorb and emit vacancies. For this model, it was shown that the steady-state creep rate is given by eq. (1) with the values of $n = 4.5$, $Q = Q_l$ where $Q_l$ is the activation energy for self-diffusion and $p = 0$ since the process occurs intragranularly. It has been shown through detailed analysis of a wide range of data for f.c.c. metals that the precise value of $n$ may be higher than 4.5 in climb-controlled creep because of a dependence upon the stacking-fault energy of the material.

For pure metals, the dislocations glide outwards very rapidly from the dislocation sources so that climb is the rate-controlling step. However, in some solid solution alloys there may be a preferential segregation of solute atoms around the moving dislocations in the form of impurity or Cottrell atmospheres and the rate of glide may then become slower than the rate of climb. Under these conditions, creep is controlled by glide in the form of a viscous drag process and it was shown that the steady-state creep rate is given by eq. (1) with $n = 3$ and $Q = Q_i$ where $Q_i$ is the activation energy for interdiffusion of the solute. Again, this process occurs intragranularly so that $p = 0$.

It has been shown, both theoretically and through detailed experiments, that the process may be even more complicated in solid solution alloys since there may be a transition with increasing stress from control by dislocation climb to control
by a viscous glide process, thereby giving a transition from
\( n = 3 \) to \( n \approx 4.5 \).\(^6\) Furthermore, at even higher stresses the
dislocations may break away from their solute atom
atmospheres giving a transition from \( n = 3 \) to \( n > 3 \).\(^7\)
Despite these potential complications, it is usually relatively
easy to identify the rate-controlling process at intermediate
and high stresses.

4. Rate-Controlling Creep Mechanisms at Low Stresses

The situation becomes more complex at low stresses, when
\( n < 3 \), because several different processes may control the
flow. Each of these processes is now considered separately.

4.1 Stress-directed vacancy flow

Diffusion creep refers to the stress-directed flow of
vacancies that occurs when there is an excess of vacancies
on those grain boundaries lying almost perpendicular to the
tensile axis and a corresponding depletion of vacancies on
those grain boundaries lying almost parallel to the tensile
axis. Vacancy flow takes place to restore an equilibrium
condition and these vacancies may flow either through the
crystalline lattice in the process known as Nabarro–Herring
(NH) diffusion creep\(^8,9\) or along the grain boundaries in the
process known as Coble diffusion creep.\(^10\) For NH creep the
rate of flow is given by eq. (1) with \( n = 1 \), \( Q = Q_0 \)
and \( p = 2 \) whereas for Coble creep the rate of flow is given by eq. (1)
with \( n = 1 \), \( Q = Q_{gb} \) and \( p = 3 \), where \( Q_{gb} \) is the activation
energy for grain boundary diffusion. Since the flow of
vacancies is equivalent to a net flow of atoms in the opposite
direction, the grains increase in length in the direction of the
tensile stress and they decrease in width in a direction
perpendicular to the tensile axis.

There is an additional, and often unrecognized, feature of
diffusion creep that must be emphasized. The elongation of
the grains within a polycrystalline matrix leads, in a very
simple way, to the lateral displacements of adjacent grains
in order to maintain coherency without the opening of voids
and this means that diffusion creep must be accommodated by
some form of grain boundary sliding. The principle of this
sliding is depicted schematically for a three-grain config-
uration in Fig. 3 where the tensile axis is vertical and the
consequence of diffusion creep on marker lines AA’ and BB’
is shown for (a) markers scribed parallel to the tensile axis
and (b) markers scribed perpendicular to the tensile axis. For
both situations, offsets are created at the points where the
markers impinge on the grain boundaries and in practice
these offsets, although a consequence of diffusion creep, are
identical in appearance to the offsets occurring due to
conventional grain boundary sliding as discussed in sec-
ction 4.2. Since the origin of this sliding is physically and
mechanistically different from conventional GBS, it is
termed Lifshitz sliding\(^11\) and it represents a situation where
there is no net increase in the number of grains measured
along the tensile axis.

4.2 Grain boundary sliding

Grain boundary sliding refers to the lateral displacement of
two adjacent grains with the movement taking place at, or
very close to, their mutual interface. Traditionally, there
always appeared to be two separate areas of GBS with sliding
occurring under conditions both of high temperature creep
and of superplasticity. More recently, however, it has been
recognized that GBS occurs in an analogous manner in both
creep and superplasticity and this provides an opportunity to
develop a unified approach to account for both situations.

Any consideration of GBS leads to the conclusion that,
when the grains have irregular shapes in the polycrystalline
matrix, sliding can occur without nucleating voids only when
it is accommodated through plastic deformation within the
adjacent grains. In practice, GBS is accommodated by the
intragranular movement of dislocations through slip but the
nature of this slip is different under creep and superplastic
conditions.

The basic mechanism of GBS is one in which sliding
occurs through the movement of dislocations along the
boundary, these dislocations generate a stress concentration
at a triple junction and/or at a grain boundary ledge, and this
stress concentration is relieved by the nucleation and move-
ment of dislocations into the adjacent grain so that the
accommodating dislocations move across the grains until
they encounter an obstacle.\(^12\) The nature of these obstacles is
dictated by the grain size. When the grain size is large as in
conventional high temperature creep, subgrains are formed
within the grains during creep and these subgrains have an
average size, \( \lambda \), which is related inversely to the level of the
applied stress.\(^3\) Thus, these subgrain boundaries represent
the obstacle to the accommodating dislocation slip under
creep conditions. When the grain size, \( d \), is smaller than the
subgrain size, \( \lambda \), no subgrains are formed and the accom-
modating dislocations then pile up against the opposing grain
boundary. It has been shown, through a detailed examination
of experimental data on two superplastic materials, that this
latter process corresponds to superplasticity\textsuperscript{13}) and the analysis is consistent with a wide range of data demonstrating that superplastic flow occurs in materials with very small grain sizes at high temperatures without the development of any intragranular subgrain structure.\textsuperscript{14)}

These two processes are illustrated schematically in Fig. 4 where (a) represents conventional creep conditions with $d > \lambda$ and the intragranular dislocations generated at A pile up at the subgrain boundary at B and (b) represents superplasticity with $d < \lambda$ and the intragranular dislocations generated at C pile up at the opposing grain boundary at D. Adopting this unified approach for creep and superplasticity, it was shown that GBS in creep in Fig. 4(a) leads to eq. (1) with $n = 3$, $Q = Q_l$ and $p = 1$ whereas GBS in superplasticity in Fig. 4(b) leads to eq. (1) with $n = 2$, $Q = Q_{gb}$ and $p = 2$.\textsuperscript{12)} Furthermore, these values of $n$, $Q$ and $p$ are consistent with experimental results showing that GBS increases in importance with decreasing stress under creep conditions\textsuperscript{15)} and they are in excellent agreement with typical experimental behavior reported for superplastic materials.\textsuperscript{14,16)} The occurrence of intragranular slip as an accommodating process in superplasticity, as illustrated in Fig. 4(b), is supported by experiments showing that dislocations become trapped in the intragranular coherent twin boundaries during deformation of a superplastic copper-based alloy\textsuperscript{17)} and the individual strains recorded in the grains of a Pb–62\%Sn alloy during superplastic flow are oscillatory in nature and make no net contribution to the total strain.\textsuperscript{18)}

When GBS occurs in the manner depicted in Fig. 4 with intragranular slip as the accommodating process, it is known as \textit{Rachinger sliding}.\textsuperscript{19)} This type of GBS differs in a very significant way from Lifshitz sliding because the grains retain their equiaxed configuration and therefore there is an increase in the number of grains measured along the tensile axis.

### 4.3 Harper–Dorn creep

The occurrence of Harper–Dorn creep as a separate and distinct creep mechanism is based on the results obtained in very early experiments conducted on pure aluminum at a high testing temperature of 920 K.\textsuperscript{20,21)} The original data are shown in Fig. 5 in a logarithmic plot of the steady-state creep rate versus the applied stress. Tests were conducted using both polycrystalline aluminum with a grain size of 3 mm and an aluminum single crystal. It is evident from inspection of Fig. 5 that the datum point obtained for the single crystal is coincident with the points obtained from the polycrystalline samples thereby demonstrating conclusively that $p = 0$ in these experiments. There is also a clear transition from $n \approx 4.5$ at the higher stresses to $n \approx 1$ at the lower stresses. The behavior at the high stresses is consistent with dislocation climb as the rate-controlling mechanism, as described in section 3, but the region with $n \approx 1$ at low stresses is not consistent with diffusion creep because there is no dependence on grain size. Furthermore, the lower broken line shows the predicted behavior for NH diffusion creep with a grain size of 3 mm and this line lies more than two orders of magnitude lower than the experimental datum points. These and other findings suggest, therefore, that Harper–Dorn creep is a separate deformation mechanism exhibiting creep behavior consistent with eq. (1) with $n = 1$, $Q = Q_l$ and $p = 0$.

Several reports are available demonstrating the occurrence of Harper–Dorn creep at low stresses\textsuperscript{22)} although questions
remain regarding impurity effects in Harper–Dorn creep\(^{23,24}\) and the role of the initial dislocation density.\(^{25}\) In addition, although the occurrence of Harper–Dorn creep is attributed to an intragranular dislocation mechanism, there are uncertainties concerning the precise nature of this process.\(^{26–31}\) There are also some reports suggesting that Harper–Dorn creep is not a genuine creep mechanism\(^{22,32}\) and it is necessary, therefore, to consider whether it is reasonable to attribute the results reported by Harper and Dorn\(^{20,21}\) to a new and unique creep process.

In the early experiments of Harper and Dorn,\(^ {20}\) transverse marker lines were scribed on the surface of a specimen in the manner depicted schematically in Fig. 3(b). This specimen was then tested at a stress of \(9.3 \times 10^{-2}\) MPa, corresponding approximately to the transition region between \(n \approx 4.5\) and \(n \approx 1\) in Fig. 5, the marker offsets were measured at 200 boundaries and the contribution from GBS to the total strain was estimated as \(\sim 12\%\). Later more detailed experiments were conducted by Harper et al.\(^ {21}\) where several specimens were scribed with transverse marker lines and the sliding contribution, \(\xi\), was measured at different levels of the applied stress, where \(\xi\) is defined as \(\xi_{gb} / \xi_t\) where \(\xi_{gb}\) and \(\xi_t\) are the strain due to sliding and the total strain, respectively. The results are plotted in Fig. 6 in terms of the sliding contribution, \(\xi\), versus the applied stress, \(\sigma\), where the vertical broken line at \(\sigma = 0.1\) MPa corresponds to the point of transition from \(n \approx 1\) to \(n \approx 4.5\) shown in Fig. 5. It is apparent in Fig. 6 that the values of \(\xi\) increase initially with decreasing stress to a maximum value of \(\xi \approx 48\%\) at \(\sigma = 2.3 \times 10^{-1}\) MPa but the values of \(\xi\) decrease at lower stress levels and they lie within the range of \(\xi \approx 5–15\%\) of the low stress region where \(n \approx 1\). As noted in section 4.1, Lifshitz sliding is required as an accommodation process in diffusion creep and Lifshitz sliding leads to offsets in marker lines which are identical to those observed in conventional GBS. Therefore, the very low values recorded for \(\xi\) in the experiments of Harper et al.\(^ {21}\) provide conclusive evidence that these results are not consistent with diffusion creep so that Harper–Dorn creep must be recognized as a separate and distinct creep process.

### Table 1 Possible creep mechanisms at low stresses.

<table>
<thead>
<tr>
<th>Creep mechanism</th>
<th>Marker offsets</th>
<th>Value of (\xi)</th>
<th>Grain elongation</th>
</tr>
</thead>
<tbody>
<tr>
<td>Diffusion creep</td>
<td>Yes</td>
<td>High</td>
<td>Yes</td>
</tr>
<tr>
<td>Lifshitz sliding</td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>Coble creep</td>
<td>Yes</td>
<td>High</td>
<td>Yes</td>
</tr>
<tr>
<td>GBS in creep ((d &gt; \lambda))</td>
<td>Yes</td>
<td>High</td>
<td>No</td>
</tr>
<tr>
<td>GBS in superplasticity ((d &lt; \lambda))</td>
<td>Yes</td>
<td>High</td>
<td>No</td>
</tr>
<tr>
<td>Intragranular dislocation creep</td>
<td>No</td>
<td>Low</td>
<td>Yes</td>
</tr>
</tbody>
</table>

### Table 2 Methods for identifying creep mechanisms at low stresses.

<table>
<thead>
<tr>
<th>Creep mechanism</th>
<th>Marker offsets</th>
<th>Value of (\xi)</th>
<th>Grain elongation</th>
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<td>Coble creep</td>
<td>Yes</td>
<td>High</td>
<td>Yes</td>
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</table>

5. Identifying the Rate-Controlling Mechanism in Creep Experiments

Creep experiments are usually conducted with the objective of identifying the rate-controlling creep mechanism for any selected conditions of temperature and stress. Generally, it is relatively easy to obtain sufficient creep data to yield an unambiguous identification of the creep mechanism or mechanisms occurring at intermediate and high stresses where creep is controlled by dislocation climb and glide. However, the situation is more complex at low stresses where there are several possible creep mechanisms.

Table 1 summarizes the creep mechanisms occurring at low stresses in terms of the values predicted for \(n\), \(p\), and \(Q\). These values are generally sufficiently similar that it is not easy to reach an unambiguous determination of the precise rate-controlling mechanism. Accordingly, it is necessary to seek additional measurements and observations that may be conducted in order to more clearly delineate the precise rate-controlling process under these conditions.

As indicated in Table 2, three additional procedures are proposed in order to achieve an unambiguous identification of the flow process. First, scribing marker lines on the specimen surfaces, either parallel or transverse to the stress axis, provides an opportunity to take detailed measurements of the marker offsets at every point where the lines cross a grain boundary. It is now well established that offsets occur in diffusion creep because of the presence of Lifshitz sliding and also in conventional GBS due to Rachinger sliding. Second, these offsets may be used to estimate the values of \(\xi\) using the standard procedures available in the literature.\(^ {34,35}\) The values of \(\xi\) will be high in diffusion creep and GBS but low in Harper–Dorn creep: an example of the low values of \(\xi\) in Harper–Dorn creep is shown in Fig. 6. Third, measurements to determine the average grain elongation provide an opportunity to distinguish between Lifshitz and Rachinger sliding.
sliding since the grains become elongated in diffusion creep but there is no grain elongation in conventional Rachinger GBS. In practice, some care must be exercised in interpreting measurements of grain elongations since it has been shown that the grain boundaries tend to migrate during creep in order to retain a reasonably equiaxed grain configuration.\(^{36}\)

These additional requirements are summarized in Table 2. Thus, by making these microstructural observations in addition to the standard measurements of strain rate recorded in creep testing, it is possible to achieve an unambiguous identification of the rate-controlling process in creep experiments conducted at low stresses.

6. Summary and Conclusions

(1) Plastic flow occurs in high temperature creep through the occurrence of three distinct types of creep mechanism: (i) intragranular dislocation creep, (ii) grain boundary sliding and (iii) diffusion creep.

(2) It is difficult to unambiguously identify the rate-controlling creep process at low stresses because of similarities between several of the possible mechanisms. For example, diffusion creep is always accommodated by Lifshitz sliding and this gives rise to offsets in surface markers which are identical in appearance to those occurring in conventional Rachinger grain boundary sliding.

(3) An alternative approach is suggested in order to achieve an unambiguous identification of the creep mechanisms at low stress levels. This approach involves measuring not only the stress exponent, \(n\), the exponent of the inverse grain size, \(1/p\), and the activation energy for creep, \(Q\), but also taking measurements of offsets in surface marker lines, determining values for the sliding contribution, \(\xi\), and measuring the average grain elongation.

Acknowledgements

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REFERENCES