CONSTRUCTING A VALIDATED DEFORMATION MECHANISMS MAP USING LOW TEMPERATURE CREEP STRAIN ACCOMMODATION PROCESSES FOR NICKEL-BASE ALLOY 718

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ABSTRACT

A creep Deformation Mechanism Map (DMM) of an engineering alloy can be an effective tool for developing physics-based prognostics systems. Many classical diffusion based rate equations have been developed for time dependent plastic flow where dislocation glide, dislocation glide-plus-climb and vacancy diffusion driven grain boundary migration (diffusion creep) are rate controlling. These creep rate equations have been proven experimentally for simple metals and alloys and form the basis of constructing an Ashby’s DMM. Long term creep testing and analysis of complex engineering alloys has shown that power law breakdown phenomenon is related to the dominance of Grain Boundary Sliding (GBS) as opposed to diffusion creep. Rate equations are now available for GBS in complex alloys and, in this paper, a DMM is constructed for a fine grained Alloy 718 and this is validated by comparison with a collection of experimental data obtained from the literature. The GBS accommodated by wedge type cracking is considered dominant at low homologous temperatures (0.3 to 0.5T_m i.e. melting temperature in Kelvin) whereas GBS accommodated by power-law or cavitation creep dominates above 0.55T_m.

NOMENCLATURE

DMM: Deformation Mechanisms Map.
GBS: Grain Boundary Sliding.
PLC: Power Law Creep.
PLB: Power Law Breakdown.
IG: Intra-Granular.
W-Type: Wedge type cracking.
R-type: Cavitation type Cracking.
ORNL: Oak Ridge National Laboratory.
IN718: Inconel Alloy 718.
4D: 4 Dimensional.
T_m: Melting Temperature in Kelvin.
T/T_m: Homologous Temperature.
σ/σ_0: Normalized Stress.
MSPE: Mean Squared Prediction Error.
INTRODUCTION

Complex engineering alloys that are used for high temperature applications such as Alloy 718, are typically precipitation-hardened and their grain boundaries are also strengthened using selective carbide precipitation mechanisms. Their grain sizes are also carefully controlled. The primary objective of using these microstructural design concepts is to optimize their high temperature yield strength as well as creep properties. Alloy 718 is widely used in the industry for high temperature applications in aircraft as well as land-based gas turbine engines. Time and temperature dependence of creep strain accumulation and formation of a creep crack are the main design criteria used in practice for this class of materials.

Practical range of creep analysis requires the availability of creep data in the strain rate range of $10^{-8}$ to $10^{-10}$/s and a temperature range of 0.3 to 0.7 $T_m$. Most academic creep studies on complex engineering alloys have focused on short term creep response because it is expedient to generate the creep data and extrapolation techniques in combination with theoretical treatments developed for simple metals and alloys have been employed to predict the long term creep response of these alloys. In contrast, long term creep studies carried out by the industry have limited their analysis using empirical modeling techniques. This divide between the academia and the industry led studies has created a lot of confusion in understanding the creep deformation phenomenon of complex engineering alloys.

EVOLUTION OF COMPLEX ALLOY DMM

The creep deformation is essentially a time dependent and thermally activated plastic strain accumulation processes that are controlled by the competition between the activation energies associated with different mechanisms such as dislocation glide and climb within the grain interiors (Power-Law Creep - PLC) or at the grain boundaries (GBS) or stress assisted vacancy migration leading to grain boundary migration (diffusion creep). All of these processes occur at the atomic scale over time and often lead to time dependent plastic strain accumulation at stresses that are well below the temperature dependent yield strength of the material.

The constitutive equations for creep define the creep strain-rate as a function of applied stress, temperature and material microstructure ($\sigma$, $T$, and $S$ in Eq. 1 where $t$ is the discretized time). It is now well recognized that the microstructure of the material itself evolves as creep strain continues to accumulate (Eq. 2) although most workers who have developed mathematical formulations for different creep deformation mechanisms tend to ignore this effect.

$$\frac{d\varepsilon}{dt} = f(\sigma, T, S)$$ (1)

Typical constitutive equations that have been developed for different creep deformation mechanisms for simple metals and alloys including some complex alloys are summarized in Table 1. Frost and Ashby [1] developed a superposition method for combining different rate equations where one mechanism contributes dominantly to the total creep rate at a given stress and temperature while other mechanisms play a less significant role. This led to the development of a DMM which is a diagram that plots normalized stress and temperature and divided the two dimensional stress and temperature space into different regions in which a specific deformation mechanism remains dominant. Different creep strain rate contours are further superimposed on this space to provide a sense of strain rates over which a specific mechanism may dominate during the deformation process. A typical DMM for Mar-M200 Ni-base superalloy with a 100 $\mu$m grain size is shown in Fig. 1, where dislocation glide, PLC and diffusion creep control the creep deformation process.

Mohammad and Langdon [2] created their own version of a DMM arguing that GBS, as opposed to Coble type diffusion creep, is likely to dominate at lower stresses because the overall energy barrier required for substantial grain boundary migration is quite large.

Koul and Castillo [3] conducted extensive creep testing on IN738LC turbine blade material with a grain size of 1.5 mm and microstructural studies on crept specimens and modified the DMM for a Ni-base superalloy by incorporating a GBS regime in a typical Ashby type DMM [4], [3], [5], shown in Fig. 2.

The transition between the PLC and GBS is obtained using experimental Power-Law Breakdown (PLB). GBS is shown by the superimposition of experimentally determined power law breakdown points in Fig. 2. This figure reveals that the GBS is the dominant mechanism within the practical range of land based and aircraft turbine operation. They further included an interface reaction controlled diffusion creep regime into the map assuming that, if diffusion creep is operative at very low stresses, the grain boundaries cannot act as perfect sources or sinks for vacancies in complex engineering alloys.

Wardsworth et. al. [6] have analyzed a vast amount of creep data that was available on different engineering alloys and concluded that GBS, as opposed to diffusion creep, was the dominant deformation mechanism in complex engineering alloys at lower stresses. A schematic of a DMM showing the dominance of PLC and inter-granular deformation is shown in Fig. 3 [7].

Recently, Banerjee et. al. [8] attempted to further subdivide the GBS dominant regime in a Ashby type DMM for a Pb-Sn eutectic solder alloy where the GBS region is further divided into
two regions based on the dominance of different GBS accommodation processes. It is suggested that GBS is accommodated by w-type cracking below 0.5Tm whereas creep cavitation accommodation process is dominant above this temperature and this modification of the DMM is presented in Fig. 4.

**FIGURE 1:** A TYPICAL DMM FOR MAR-M200 WITH GRAIN SIZE OF 100 µm [1].

**FIGURE 2:** A MODIFIED DMM FOR IN 738LC WITH GRAIN SIZE OF 1.5 m [3].

**FIGURE 3:** A SCHEMATIC CREEP MECHANISMS. [7].

**FIGURE 4:** A MODIFIED DMM WITH GBS USED AS THE TEMPLATE OF THIS PAPER. [8].

**CONSTRUCTING A MODIFIED DEFORMATION MECHANISM MAP (DMM)**

The construction of a practical DMM using numerical techniques requires the availability of long as well as short term creep test data in order to obtain the correct numerical output with respect to the experimental data over a wide range of temperatures and stress conditions. As pointed out earlier, it is not sufficient to perform creep tests at higher temperatures alone where creep rates are relatively higher and use some theoretical assumptions to construct the map. As a minimum, some test data should be available in the PLC and PLB regimes to capture the transition between the PLC and the GBS. The practical range of temperatures and stresses in the case of cast turbine blades is limited to a


<table>
<thead>
<tr>
<th>Mechanisms</th>
<th>Rate Equation $[\varepsilon^{-1}]$</th>
<th>Remarks</th>
</tr>
</thead>
<tbody>
<tr>
<td><strong>Dislocation Climb</strong></td>
<td>$\dot{\varepsilon} = AD\dot{G}b (\sigma / G)^{4.5}$</td>
<td>$A \approx 10^6$ $G =$ shear Modulus</td>
</tr>
<tr>
<td><strong>Diffusional Creep</strong></td>
<td></td>
<td></td>
</tr>
<tr>
<td>Nabarro-Herring (Bulk)</td>
<td>$\dot{\varepsilon} = BD\dot{G}b (\sigma / G)^2$</td>
<td>$B \approx 30$ $G =$ shear Modulus $\sigma =$ Threshold Stress $\geq 10^{-7}G$</td>
</tr>
<tr>
<td>Coble (Grain Boundary)</td>
<td>$\dot{\varepsilon} = CD (\dot{D_{c}}Gb) b^2 (\sigma / G)$</td>
<td>$C \approx 30$ $G =$ shear Modulus $\sigma =$ Threshold Stress $\geq G_b^2$</td>
</tr>
<tr>
<td>Ashby-Verrall (Diffusional Accommodated Flow)</td>
<td>$\dot{\varepsilon} = FD\dot{G}b (\sigma - \sigma_1)[1 + \frac{\delta}{\Gamma} (\dot{D_{c}}Gb)]$</td>
<td>$F \approx 100$ $G =$ shear Modulus $\sigma =$ Threshold Stress $= \sigma_1 + \frac{\delta}{\Gamma}$</td>
</tr>
<tr>
<td><strong>Grain Boundary Sliding</strong></td>
<td></td>
<td></td>
</tr>
<tr>
<td>Controlled by GB Diffusion</td>
<td>$\dot{\varepsilon} = HDgb (\sigma / G)^2$</td>
<td>$H \approx 8 \times 10^5$ $G =$ shear Modulus</td>
</tr>
<tr>
<td>Controlled by Lattice Diffusion</td>
<td>$\dot{\varepsilon} = LD\dot{G}b (\sigma / G)^2$</td>
<td>$L \approx 10^7$ $G =$ shear Modulus</td>
</tr>
<tr>
<td><strong>Ashby-Verrall</strong></td>
<td>$\dot{\varepsilon} = MD\dot{G}b (\sigma / G)$</td>
<td>$M \approx 10^{-11}$ $G =$ shear Modulus</td>
</tr>
<tr>
<td>Complex Engineering Alloy</td>
<td>$\dot{\varepsilon} = AD\dot{G}b (\sigma / G)^{q-1} (\sigma - \sigma_{\text{climb}}) G^2 / (G_b^2)$</td>
<td>$A \approx 10^6$ $G =$ shear Modulus $q =$ 1 (without particles) $q =$ 3 (continuous network of particles) $q =$ 4 (discrete particles) $G =$ GB free energy $b =$ interedge spacing $r =$ average grain boundary ledge height</td>
</tr>
</tbody>
</table>

**TABLE 1: VARIETY OF CREEP MECHANISMS AND FORMULA [9], Complex Eng. Alloy [10].**

As pointed out by Wardsworth et al. [6], in complex engineering alloys, intense defect activity in the grain boundary plane and the adjacent regions during GBS takes place and strain accommodation processes in the form of precipitate denudation zones, minor migration and grain boundary cracking also remain active.

In the case of wrought alloys such as Alloy 718, it has been suggested that w-type cracking at the grain boundaries and grain boundary cavitation (r-type) are the main GBS strain accommodation processes [8], [11]. However, the possibility of a transition between the w-type and r-type strain accommodation processes as a function of stress and temperature has not been explored. There are currently only few (in some case none) experimental data bases available in the literature on creep tests at the relatively lower temperatures ($0.5T_m$ or below) to observe the transition between the GBS accommodated by w-type and r-type cracking. Fig. 5 shows two fracture surfaces for wrought Ni-base superalloys; one very close to the $0.5T_m$ where w-type cracking occurs and the other above the $0.55T_m$ where grain boundary cavitation is dominant.

**DATA COLLECTION**

Alloy 718 was selected as a test case because a fair amount of creep information is available in the literature for this alloy. An extensive literature survey was conducted to collect as much creep data as possible. The data collection exercise focused on
the minimum creep rate which is the most critical design parameter that can be extracted from a constant load creep test from a life prediction perspective [12]. Values for creep strain rates were plotted for tests performed on samples with a grain size of approximately 30 µm. In cases where grain size information was not available, it was assumed that the samples had received the conventional heat-treatment and the grain size lay in the range of 30 µm to 50 µm [13]. Eq. 3 was used to normalize the effect of grain size on the creep strain rate if the grain size was not in this range. In this equation, $d_1$ and $d_2$ denote grain sizes for two different creep rates. A schematic representation of the grain size effect is shown in Fig. 6 [9]. This plot indicates that a careful use of Eq. 3 is necessary when grain size is larger than 300 to 400 µm. However, the grain sizes used in this paper lay on the left hand side of the minimum where the Eq. 3 can still be applied. Since this equation lies in the GBS regime, the creep rates were not adjusted for creep experimental points in the dislocation glide regime. Generally, each deformation mechanism leads to a different value of the stress exponent (n) in the Arrhenius-type creep equation [12]. Thus, this value can be used to identify the stress range in which a certain mechanism is dominant.

$$\dot{\varepsilon}_2 = \dot{\varepsilon}_1 \left( \frac{d_1}{d_2} \right)^2$$

Eq. 3

Fig. 7 shows a collection of experimentally measured minimum creep rates at Oak Ridge National Laboratory (ORNL) for fine granular Alloy 718 [14]. Table 2 provides the values extracted from this plot. The minimum creep rate values are converted to [hr$^{-1}$] to conform with the computational results calculated in this paper.

The temperature dependence of strain rates at two different stress levels (254, 354 MPa) reported for Alloy 718 indicates a break-down at about 0.55$T_m$ (where, $T_m \sim 1600 °K$ is the melting point of the alloy) as shown in Figure 8. Since a polynomial might be employed to fit the data, such a break-down is hard to recognize in common log-plot is used for creep analysis. This break-down is consistent with the w-type cracking observed in Udimet 520 and r-type cavitation in Waspaloy at temperatures below and in excess of 0.55$T_m$, respectively.

**COMPUTATIONAL MODEL FOR DMM**

A frequently-used model for creep strain rate around the PLB is the superposition of two strains that are grain boundary sliding and intragranular (IG) deformation (or dislocation glide-plus-climb) as shown in Fig. 9. There is a variety of rate equa-
Homologous Temperature | Normalized Shear Stress | Minimum Creep Rate
<table>
<thead>
<tr>
<th></th>
<th></th>
<th></th>
</tr>
</thead>
<tbody>
<tr>
<td>$T / T_m$ [°K]</td>
<td>$\log_{10}(\sigma/G_0)$ [MPa]</td>
<td>$[hr^{-1}]$</td>
</tr>
<tr>
<td>0.61</td>
<td>-2.60</td>
<td>$6.79 \times 10^{-5}$</td>
</tr>
<tr>
<td>0.61</td>
<td>-2.53</td>
<td>$1.40 \times 10^{-4}$</td>
</tr>
<tr>
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<td>-2.48</td>
<td>$2.11 \times 10^{-4}$</td>
</tr>
<tr>
<td>0.61</td>
<td>-2.38</td>
<td>$1.92 \times 10^{-4}$</td>
</tr>
<tr>
<td>0.61</td>
<td>-2.34</td>
<td>$6.37 \times 10^{-4}$</td>
</tr>
<tr>
<td>0.61</td>
<td>-2.31</td>
<td>$5.43 \times 10^{-4}$</td>
</tr>
<tr>
<td>0.61</td>
<td>-2.24</td>
<td>$2.81 \times 10^{-3}$</td>
</tr>
<tr>
<td>0.61</td>
<td>-2.18</td>
<td>$3.28 \times 10^{-3}$</td>
</tr>
<tr>
<td>0.61</td>
<td>-2.15</td>
<td>$1.45 \times 10^{-2}$</td>
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<td>-2.60</td>
<td>$6.14 \times 10^{-6}$</td>
</tr>
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<td>0.58</td>
<td>-2.48</td>
<td>$1.19 \times 10^{-5}$</td>
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<td>0.58</td>
<td>-2.38</td>
<td>$2.03 \times 10^{-5}$</td>
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<td>-2.27</td>
<td>$5.39 \times 10^{-5}$</td>
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<td>-2.24</td>
<td>$1.05 \times 10^{-4}$</td>
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<td>$5.70 \times 10^{-3}$</td>
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<td>-2.21</td>
<td>$2.38 \times 10^{-2}$</td>
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<td>0.58</td>
<td>-2.18</td>
<td>$3.94 \times 10^{-2}$</td>
</tr>
<tr>
<td>0.58</td>
<td>-2.17</td>
<td>$5.94 \times 10^{-2}$</td>
</tr>
<tr>
<td>0.58</td>
<td>-2.12</td>
<td>$6.95 \times 10^{-2}$</td>
</tr>
<tr>
<td>0.54</td>
<td>-2.24</td>
<td>$5.52 \times 10^{-6}$</td>
</tr>
<tr>
<td>0.54</td>
<td>-2.18</td>
<td>$7.10 \times 10^{-6}$</td>
</tr>
<tr>
<td>0.54</td>
<td>-2.13</td>
<td>$8.84 \times 10^{-6}$</td>
</tr>
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<td>0.54</td>
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<td>$2.43 \times 10^{-5}$</td>
</tr>
<tr>
<td>0.54</td>
<td>-2.04</td>
<td>$7.34 \times 10^{-5}$</td>
</tr>
<tr>
<td>0.54</td>
<td>-2.01</td>
<td>$2.15 \times 10^{-4}$</td>
</tr>
</tbody>
</table>

**TABLE 2:** Minimum Creep Rate Experimentally Measured and Reported by ORNL for Inconel 718.

Torsions for grain boundary sliding and intragranular strain rates and, in this paper, the rate Eqs. 5 and 6 were initially used for GBS and IG (from Table1). The constant coefficient and power values have then been modified to obtain a better agreement with the experimental data. The term $\sigma$ in intragranular rate equation is the effective stress experienced by the dislocations and this is altered in [9] to $\sigma - \sigma_0$ where $\sigma_0$ is the friction stress or the internal stress inside the material, that is, the minimum stress required for dislocations motion process in a precipitation hardened alloy. The term $\sigma_0$ is associated with precipitation hardening [9] and set constant here because the precipitation’s size and distribution are assumed unchanged during the creep process. The computational algorithm has been developed using the Open-source R-Project

\[ \dot{\varepsilon} = \dot{\varepsilon}_{\text{gbs}} + \dot{\varepsilon}_{\text{ig}} \]  

\[ \dot{\varepsilon}_{\text{gbs}} = \frac{H D_{\text{gbs}} G b}{kT} \left( \frac{b}{d} \right)^2 \left( \frac{\sigma}{G} \right)^2 \]  

\[ \dot{\varepsilon}_{\text{ig}} = \frac{A D_{\text{ig}} G b}{kT} \left( \frac{\sigma - \sigma_0}{G} \right)^{4.5} \]
Table 3 shows the values employed for computationally developing the DMM including the references used in each case. Temperature-dependent diffusion coefficients (for both GBS and IG deformation) and shear modulus are used and updated in the rate equations for given temperatures. The relationships are given in Eqs. 7 [1], 8 [1], and 9 [17]. The temperature-dependent shear modulus equation is based on experimental measurements conducted by Farraro [17] for Young’s modulus and converted to shear modulus using \( \mu = E/(1 + \nu) \) where \( \mu, E, \) and \( \nu \) are shear modulus, Young’s modulus, and Poisson ratio equal to 1/3 respectively.

### TABLE 3: VALUES EMPLOYED IN COMPUTATION ALGORITHM [9].

<table>
<thead>
<tr>
<th>Parameter</th>
<th>Notation</th>
<th>Value</th>
<th>Unit</th>
<th>Reference</th>
</tr>
</thead>
<tbody>
<tr>
<td>Melting Temperature</td>
<td>( T_m )</td>
<td>1600 ( ^\circ )K</td>
<td></td>
<td>[16]</td>
</tr>
<tr>
<td>Burger Vector</td>
<td>( b )</td>
<td>( 2.5 \times 10^{-4} ) ( \mu m )</td>
<td></td>
<td>[1]</td>
</tr>
<tr>
<td>Boltzmann Constant</td>
<td>( k )</td>
<td>( 1.38 \times 10^{23} ) ( J/\mu m^k )</td>
<td></td>
<td>[1]</td>
</tr>
<tr>
<td>Pre-exponent, Intragranular</td>
<td>( D_{gb} )</td>
<td>( 2 \times 10^{-6} ) ( m^2/s )</td>
<td></td>
<td>[1]</td>
</tr>
<tr>
<td>Activation Energy, Intragranular</td>
<td>( Q_i )</td>
<td>425000 ( J/mol )</td>
<td></td>
<td>[14]</td>
</tr>
<tr>
<td>Gas Constant</td>
<td>( R )</td>
<td>8.3144 ( J/mol^k )</td>
<td></td>
<td>[1]</td>
</tr>
<tr>
<td>Grain Size</td>
<td>( d )</td>
<td>30 ( \mu m )</td>
<td></td>
<td>[18]</td>
</tr>
<tr>
<td>Friction Stress</td>
<td>( \sigma_0 )</td>
<td>( 22 \times 10^6 ) ( \mu )</td>
<td></td>
<td>[9]</td>
</tr>
<tr>
<td>Pre-exponent, gbs</td>
<td>( D_{gb} )</td>
<td>( 46^2 D_{gb} ) ( m^2/s )</td>
<td></td>
<td>[1]</td>
</tr>
<tr>
<td>Activation Energy, gbs</td>
<td>( Q_s )</td>
<td>0.6( Q_e ) ( J/mol )</td>
<td></td>
<td>[1]</td>
</tr>
</tbody>
</table>

\[
D_{ig} = D_{gb} \exp \left( \frac{-Q_i}{RT} \right) \quad (7)
\]

\[
D_{gbs} = D_{gb} \exp \left( \frac{-Q_s}{RT} \right) \quad (8)
\]

\[
G = (0.93 - 3.59 \times 10^{-4} T) \times 10^{11} \text{ [Pa]} \quad (9)
\]

### COMPUTATIONAL RESULTS AND EXPERIMENTAL DATA

The results of the predicted DMM are shown in Fig. 11 and the experimental data from Table 2 are added to the plot as triangle symbol for comparison. The Y-axis is log-plot and the values are \( \log(\sigma_{gb}) \). For example, the value -2.2 corresponds to 523 [MPa] calculated by normalized shear stress \( \sigma = 0.0063 \times 83000 \text{MPa} = 523 \text{MPa} \) \( G_0 \) is the shear modulus at 273 \( ^\circ \)K which is about 83,000 [MPa]. The points on the map are the rates given in Table 2. Adding the experimental rate values beside the triangles on this plot makes it inconvenient to read. Readers may use Table 2 for the experimental values. The error in experimental measurement that can be assessed by repeating the tests requires for a valid comparison. Although a typical creep test includes some experimental uncertainties, repeating the creep test is not common because the test takes too long. It should be mentioned that the computational results is deterministic and is comparable to the trend lines in Fig. 7 rather than the data points.

The slope of the trend lines added to the ORNL experimental data (log-plot in Fig. 7) suggests a modification to the n-power values in rate equations such that the power changes from 4.5 to 7 for the intragranular rate equation (dislocation climb). The constant coefficient has also changed for these equations from \( 10^6 \) to \( 8 \times 10^{15} \) for the intragranular rate equation and from \( 8 \times 10^5 \) to \( 10^10 \) for the grain boundary sliding rate equation in order to develop a better prediction compared to the experimental data. The boundary between two regions; GBS and PLC is plotted presenting the transition from one mechanism to the other one. Further in this paper, the method to define this boundary has been explained in details. The alloys 718 is usually strengthened by \( \gamma' \) and \( \gamma'' \) precipitates. Fig. 10 presents \( \gamma' - \text{solvus} \) temperature for different superalloys and this is about 1030 °C for alloy 718. Above the \( \gamma' - \text{solvus} \) temperature, the basic nature of material completely changes due to dissolution of the major strengthening phase and therefore the rate equations can not be used. The vertical line at 0.82 homologous temperature on presented DMM shows the \( \gamma' - \text{solvus} \) temperature.

Fig. 12 presents a 4D contour plot that visualizes two rate contours together. This helps understand the contribution from each mechanism to the total strain rate shown in Fig. 11 as well shows the transition boundary between the two mechanisms. The intersections between the black contours (GBS rate) and color contours (interagranular/climb) demonstrate the transition boundary between the two mechanisms where the dominant mechanism is switched from one to another. For example, if one follows the strain rate contour \( 10^{-6} \), \( 10^{-4} \), and \( 10^{-2} \) on both GBS and IG contours, finds the intersections and connects the intersection points together, this defines the transition boundary on which the GBS and PLC mechanisms are switched. This is based on the physical fact that the strain rates become equal on the transition boundary. A solid blue line that connects three circular points is added to this plot and shows the experimentally defined boundary from Fig 7.

Deviation of computationally developed results from the experimental data points is measured by using the statistical Mean
Figure 10: Relation between γ' volum fraction and γ' - solvus temperature for Ni-base superalloy [5].

Squared Prediction Error (MSPE) formulation given in Eq. 10 where \( \hat{P} \) is the set of our computational predictor, \( P \) is the set of experimental values, and \( n \) is our population size i.e. 26 data points (Table 2). The MSPE is 0.0004 for this analysis. The MSPE measures the expected (or the most probable) squared distance between what our computational predictor predicts for a specific stress and temperature is and what the actual experimental value is. Creep analysis has large variation in experimental measurement and it is not subtracted from the calculated MSPE here.

\[
MSPE(P) = \frac{1}{n} \sum_{i=1}^{n} [(P(x_i) - \hat{P}(x_i))^2] \tag{10}
\]

Conclusion
A rationale and a methodology for developing practical Ashby type DMM has been created and summarized in this paper. Unlike in traditional Ashby type DMM, where diffusion creep is considered dominant at lower stresses and temperatures, GBS is considered to be the dominant deformation mechanisms below the PLB point as a function of stress and temperature. The GBS regime is further subdivided into w-type and r-type cracking accommodation processes for wrought alloys.

A comparison between the modified Ashby type DMM and experimental data obtained from the literature indicates that Eq. 11 along with Eq. 7, 8, and 9 can accurately predict the creep rates for Alloy718.
\[ \dot{\varepsilon} = \dot{\varepsilon}_{gbs} + \dot{\varepsilon}_{ig} \]

\[ \dot{\varepsilon}_{gbs} = 10^{10} D_{gbs} \frac{Gb}{kT} \left( \frac{b}{d} \right)^{2} \left( \frac{\sigma}{G} \right)^{2} \]

\[ \dot{\varepsilon}_{ig} = 8 \times 10^{13} D_{ig} \frac{Gb}{kT} \left( \frac{\sigma - \sigma_{0}}{G} \right)^{7} \] (11)

Although a grain size term appears in this relationship, the effect of grain size was not compared to the experimental data and this relationship is valid for grain sizes ranging between 30 to 50 µm. Further work will be required to study the effect of grain size.

The experimental data at low temperature (below 0.55Tm) and high stress shows that the power values tends to increase, for example from 7 to 11, in the grain boundary sliding rate equation. One may modify the equation for better prediction over this region.

This DMM can be used for creep life predictions of alloy 718 in ranges of temperatures and stresses used in practical applications, as well as to design experimental creep test matrices for wider ranges of stresses and temperatures.

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**REFERENCES**


