Creep properties of an extruded copper–8% chromium–4% niobium alloy

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Abstract
Constant-stress creep experiments were conducted on extruded Cu–8 Cr–4 Nb (GRCop-84) from 0.57 to 0.79 Tm, with observed strain rates spanning over four orders of magnitude. GRCop-84 is composed of approximately 14 vol.% Cr2Nb distributed in sizes ranging from approximately 30 nm to 0.5 μm in a matrix of pure copper with a grain size of approximately 1–3 μm. This high-strength alloy exhibits creep curves with a standard pure metal-type appearance and true failure strains typically in excess of 10%. The Monkman–Grant relationship is obeyed with a slope of −1.08. Computation of activation energies between neighboring isotherms yields an average value of 287 ± 14 kJ/mol. Despite apparent power law creep behavior within an isotherm, a master plot of temperature compensated strain rate (˙ε/ε0 exp[−Q/RT]) versus temperature compensated stress (σ/ε0 G) demonstrates that most of the data are above power law breakdown, consistent with σ/G > 10−3. A hyperbolic sine model facilitates interpolation on this master plot to stresses and/or temperatures where experimental data do not exist. Neither the threshold stress model nor the dislocation detachment model applies to GRCop-84 under these creep conditions.

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1. Introduction
Many engineering applications require the use of creep resistant materials. Of particular concern in this work is the creep performance of a particle-reinforced copper alloy being considered for the combustion chamber liner and nozzle of the next generation space shuttle main engine (SSME). These components will be actively cooled, requiring high thermal conductivity and good creep resistance. This application requires a relatively pure matrix with the addition of non-shearable particles to improve creep strength to suitable levels.

One material that has been proposed for this application is a particle-reinforced alloy developed by NASA Glenn Research Center at Lewis Field (NASA GRC), known as GRCop-84. This alloy has a fine-grained, pure copper matrix with approximately 14 vol.% particulate obtained through extruding argon atomized powders [1]. The particulate reinforcements in this alloy are Cr2Nb that occur as approximately 30–500 nm diameter particles [2–4]. The copper grain size is reported to be between 1 and 3 μm in the as-extruded material [1,3].

GRCop-84 is a difficult material to categorize as being either dispersion strengthened (DS), precipitation hardened (PH), or a metal matrix composite (MMC). It is unlike traditional PH alloys in that the particles form during rapid solidification rather than by precipitating from a solid solution. GRCop-84 is similar to a MMC in that it has a relatively high volume fraction reinforcement with the largest reinforcement size similar to the largest grain size. These particles are too large to interact with moving dislocations and probably improve strength by a load transfer mechanism. The alloy is also similar to DS materials since Cr2Nb is also present as small (few tens of nanometers) intragranular particulate.

The creep behavior of DS and MMC materials is poorly characterized by the power law model relationship that is normally used to represent the dependence of creep strain rate on stress in pure metals and some metal alloys [5].
stress exponents in particle-strengthened materials are often very large (20–100 [6,7]) and the activation energies are significantly larger than those for matrix self-diffusion [6]. These limitations of the power law model have led to numerous efforts to formulate improved deformation behavior models in DS and MMC materials. Most of these models are phenomenological in nature, but a few have a fundamental basis. While a single unifying model of creep behavior in DS and MMC materials does not exist, two prominent models, the first phenomenological and the second fundamental, are considered in this paper and they are discussed below.

2. Threshold stress model

The basis of the threshold stress model is that the applied stress is reduced by the so-called threshold stress, resulting in an effective stress that drives dislocations over particles. A form of this model was proposed by Barrett [8] and is used in this paper written as:

\[ \dot{\varepsilon} = A \times \left( \frac{\sigma - \sigma_0}{G} \right)^n \]  

where \( \dot{\varepsilon} \) is the observed steady-state strain rate, \( A \) the material constant, \( \sigma \) the applied stress, \( \sigma_0 \) the threshold stress, \( G \) is the temperature-dependent matrix shear modulus, and \( n \) is the stress exponent. This model has been used to describe a number of materials regardless of whether the creep data exhibit curvature on a strain rate–stress plot, indicating the presence of a threshold (e.g., Nardone and Strafe [9]). Often investigators invoke this model as a way to reduce the apparent values of the stress exponent and activation energy to magnitudes that can be rationalized. The threshold stress model has been applied to a wide variety of materials [7], notably rapidly solidified aluminum alloys [10], oxide DS nickel alloys [11,12], particle strengthened intermetallics [13], and powder processed aluminum alloys [14].

The threshold stress model has at least three notable shortcomings. While this model does effectively describe some experimental data. In addition, both the interaction parameter and the detachment stress depend sensitively on the difference in interaction parameters \( Q_{app} \) and \( Q \).

The RA model has three significant shortcomings. Dislocation density is assumed constant and often must be made unrealistically large to correlate model predictions to experimental data. In addition, both the interaction parameter and the detachment stress depend sensitively on the difference in interaction parameters \( Q_{app} \) and \( Q \).

The RA creep equation only applies for stresses less than the detachment stress. Since most strengthening materials have service temperatures above zero Kelvin, creep can occur with thermal assistance for applied stresses less than the detachment stress. Rösl er and Arzt demonstrate that localized climb can be stabilized if the dislocation line energy is relaxed at the particle/matrix interface [16]. The magnitude of the detachment stress is proportional to the extent of relaxation described by the dimensionless relaxation parameter \( k \), as defined in their paper:

\[ \sigma_d = \sigma_{0a} \sqrt{1 - k^2} \]  

where \( \sigma_{0a} \) is the Orowan stress and is computed by:

\[ \sigma_{0a} = 0.84 \frac{MGb}{\pi(2d - 2\sqrt{1 - T} \ln \frac{r}{b})} \]  

where \( r \) is the average particle radius, \( 2d \) is the interparticle spacing, \( M \) is the Taylor factor, \( v \) is Poisson’s ratio, \( b \) is the magnitude of the Burgers vector, and other terms are as previously defined. The relaxation parameter can, in principal, vary from 0 to 1, but as a practical matter it is usually in the range 0.77 \( \leq k \leq 0.94 \) [6]. The creep rate is given by:

\[ \dot{\varepsilon} = \dot{\varepsilon}_s \exp \left[ \frac{Q_{app} - Q}{kT} \right] \]  

where \( \dot{\varepsilon}_s = 6D_{app} \lambda \mu / b \) is a reference strain rate, \( \mu \) is the shear modulus, \( Q_{app} \) is the activation energy for vacancy diffusion in the pure matrix material \( Q \).

The RA model has three significant shortcomings. Dislocation density is assumed constant and often must be made unrealistically large to correlate model predictions to experimental data. In addition, both the interaction parameter and the detachment stress depend sensitively on the difference in interaction parameters \( Q_{app} \) and \( Q \).

The RA creep equation only applies for stresses less than the detachment stress. As such, the detachment stress represents an upper limit for the creep stress, whereas the threshold stress represents a lower limit above which creep can occur. The detachment stress and threshold stress should not be equated unless a transition between these two mechanisms exists.

3.1. Objectives

The objective of this work is to characterize the tensile creep properties of extruded GRCop-84 tested under isothermal, isostress conditions from 773 to 1073 K. A simple power law model, as well as the two models presented in
detail above are applied to gain understanding of deformation mechanisms in this material. In addition, we provide a preliminary comparison of the creep properties of this material in both the rolled and extruded conditions.

4. Experimental methods

4.1. Material

GRCop-84 is manufactured as both extruded bar stock and rolled sheet (rolled from extruded material). Examination of polished and etched, as-received extruded material shows a non-uniform distribution of particles (Fig. 1). Extruded GRCop-84 exhibits regions sparsely occupied by large particles or particle agglomerations and densely populated regions of small particles. The latter appear to occur in bands parallel to the extrusion axis and the matrix grain size is smaller in these regions. The matrix grain size is larger in those regions with large particles or particle agglomerations.

4.2. Specimen preparation

Three extruded bars of GRCop-84 were obtained from NASA GRC. These bars had a steel skin surrounding a GRCop-84 core, with a core diameter of ~2.5 cm, and were manufactured by filling a steel can with powdered GRCop-84 and extruding at ~1133 K, with a minimum area reduction of 8:1 required for full consolidation [17]. Flat tensile specimens were wire electro-discharge machine (wEDM) cut on all surfaces. The specimens were fabricated by first cutting them into rectangular sheets 0.040 in. thick, then cutting the “dogbone” profile, and finally stacking several specimens together to drill holes in each end. A similar wEDM procedure has previously been used on GlidCop Al-15 [18].

A sheet of approximately 1.19 mm thick pre-production rolled GRCop-84 material was also obtained from NASA GRC. This material has been called pre-production since it was not rolled using tension-tension methods, unlike later production rolled material that was rolled using the standard tension-tension technique [17]. This material was rolled from previously extruded bar stock followed by a 15% warm roll per pass from 5.84 mm thick at 523 K. The process was completed with a 60-min stress relief anneal at 523 K. Test specimens were cut to final dimensions from the sheet using wEDM.

Following wEDM cutting all specimens were bead blasted using very fine glass beads. Bead blast and recast wEDM surface effects were partially removed from extruded specimen faces by dry sanding. Sanding was done in a fixture by hand on 600 grit SiC paper, which removed an average of approximately 43 μm in thickness and 13 μm along the edges of the gage section. All specimens were acetone washed prior to creep testing.

4.3. Creep testing procedure

All creep experiments were conducted in tension on a constant stress creep machine under vacuum (pressure ≤ 2 × 10⁻⁵ Torr). Constant stress was achieved through the combination of specimen design and a 3:1 Andrade-Chalmers (AC) cam [19]. Temperature was monitored using an Inconel-sheathed Type K thermocouple in situ. Heating was...
initiated once vacuum fell below $5 \times 10^{-5}$ Torr. A total of 3 h elapsed from initiation of heating to application of the mass used to begin creep testing. All specimens were furnace cooled post fracture (~2 h to reach room temperature).

Displacement of the load train was monitored using a Schaevitz 300 HR Linear Variable Differential Transformer coupled to a Dasytrol 3130 signal conditioner. Both the conditioned LVDT signal and the millivolt level thermocouple were digitized via a Hewlett-Packard (HP) 3497A multiplexing multimeter resulting in a strain resolution of 3.3 $\times$ 10$^{-6}$ mm/mm and a temperature resolution of ±1.2 K. Custom software was used to record data at user selectable intervals.

Specimens were creep tested at five nominal isotherms: 773 K (0.57 $T_0$), 823, 923, 1023, and 1073 K (0.79 $T_0$). This spacing was chosen to facilitate determination of creep activation energy. The results are presented in terms of the measured average temperatures rather than the nominal temperatures. Applied stresses varied within each isotherm, but in general creep testing was limited at high stress by the apparent onset of power law breakdown and at low stress by the time to rupture. Overall, stresses ranged from 12.5 MPa (1073 K) to 200 MPa (773 K). The longest time to failure was just less than 1600 h.

5. Parametric description of creep results

True strain versus time plots of GRCop-84 have the appearance of classic creep curves for pure metals with distinct primary, secondary (steady-state), and tertiary regions (Fig. 2). Plots of true strain rate versus true strain demonstrate that the secondary creep rate in extruded GRCop-84 is constant for a range of one to two percent true strain. This behavior contrasts with some particle strengthened materials where the secondary creep rate is constant for a range of one to two percent true strain. This appearance of classic creep curves for pure metals with dislocation stress fields. The temperature dependence of creep rate can be obtained taking the natural logarithm of Eq. (5) and then the partial derivative with respect to either $T$ or $1/T$.

$$\dot{\varepsilon} = A \exp \left( -\frac{Q}{RT} \right) \times \left( \frac{\sigma}{G} \right)^n$$

(5)

where $Q$ is the activation energy, $R$ is the universal gas constant, $T$ is the temperature, and the other terms are as previously defined. Activation energy can be determined by taking the natural logarithm of Eq. (5) and then the partial derivative with respect to either $T$ or $1/T$. The resultant expressions are:

$$Q = R T \left( \frac{\partial \ln \dot{\varepsilon}}{\partial T} \right)_{\text{constant}} \approx RT^2 \frac{\Delta \ln \dot{\varepsilon}}{\Delta T} \text{constant}$$

(6a)

or

$$Q = -R \left( \frac{\partial \ln \dot{\varepsilon}}{\partial (1/T)} \right)_{\text{constant}} \approx -R \frac{\Delta \ln \dot{\varepsilon}}{\Delta (1/T)} \text{constant}$$

(6b)

To compute activation energy, one must compare strain rate data from neighboring isotherms at constant $\sigma/G$. Due to the high stress dependence of strain rate in GRCop-84, it is difficult to generate experimental data that fully conform to this requirement. An analytical method was developed instead. A range of $\sigma/G$ within the power law region common to two adjacent isotherms was identified and, within this range, 10 equally spaced values of $\sigma/G$ were identified. For each $\sigma/G$ value, an interpolated value of $\ln \dot{\varepsilon}$ was computed using the power law fit values of $A$ and $n$ for the appropriate isotherm.

5.1. Extruded GRCop-84

Fig. 3 is a double-logarithmic plot of strain rate versus stress for extruded GRCop-84. At a nominal temperature of 1068 K, the data reveal a bilinear behavior with $n = 8.5$ at
The S.E. and 95% CI are presented for the power law parameters.

The average creep activation energy for extruded GRCop-84 is larger than the expected value of activation energy—namely that for self-diffusion in pure copper (197 kJ/mol, [21]). The activation energy in extruded GRCop-84 is also larger than the activation energy for creep in pure copper when compared at similar temperatures [22–24]. The magnitudes of the stress exponents (except for the low stresses at 1068 K) are marginally too large to directly identify a creep mechanism that is consistent with the simple power law model, but these results are typical of DS or MMC metals.

### 5.2. Pre-production rolled GRCop-84

Pre-production rolled GRCop-84 creeps faster than extruded material at the same temperature and applied stress (Fig. 3). At 764 K, pre-production rolled GRCop-84 has a stress exponent of 7.6 and has a creep rate approximately 10 times faster than material in the extruded condition with a stress exponent of 8.7. Very limited data exists at 917 K which suggests that pre-production rolled material has a stress exponent of 10 and has a creep rate approximately 11 times faster than extruded GRCop-84 with a stress exponent of 8.0. The mechanisms responsible for these differences are being explored in a more thorough study of rolled material.

### 5.3. Threshold stress model

In Eq. (1), the quantity $\sigma - \sigma_o$ represents an effective stress or the thermal component of the stress available for creep, where $\sigma_o$ signifies the athermal component of the flow stress. Creep will only occur when the applied stress is greater than the threshold stress. Most investigators adopting this model to describe creep data for reinforced matrix materials use one of several stress exponents based on known deformation mechanisms in the unreinforced matrix material. Threshold stress is then determined as the intercept
on a linear–linear plot of $\dot{\varepsilon}^{1/n}$ versus $\sigma$, as introduced by Lagneborg and Bergman [25].

The threshold stress model was applied to extruded and pre-production rolled GRCop-84, but not using the standard methods. The usual approach taken to compute the threshold stress requires first assuming a stress exponent and subsequently applying the Lagneborg-Bergman [25] method described above. While this approach is widely utilized, it has a serious deficit in that the stress exponent must be divined prior to analyzing creep data. The choice of stress exponent naturally has an effect on the magnitudes of threshold stress and activation energy determined.

Rather than assuming a particular stress exponent, a nonlinear fit calculated with the threshold stress model was applied to GRCop-84 creep data using the Nonlinear Regression function within the Nonlinear Statistics package of Mathematica. The Levenberg-Marquardt minimization method was chosen to minimize the $\chi^2$ merit function. All data within an isotherm and a condition (i.e., pre-production rolled or extruded) were fit to Eq. (1) by taking $\sigma_o$ to be a temperature-dependent constant or by taking $\sigma_o$ to scale with the applied stress as $B\sigma$, where $B$ is a real-valued, positive constant. In both cases, data were provided as triplets of stress, shear modulus, and strain rate. Shear modulus was computed using the expression given by Frost and Ashby [21] at the time-weighted average temperature during secondary creep. Both models were evaluated as untransformed data and as natural logarithm-transformed data. In the latter case, the natural logarithm was taken of both strain rate and of $\dot{\varepsilon}$. All models had three fit parameters: $A$, $\sigma_o$ or $B$, and $n$.

Proper initial values that would lead to convergence of the fit routine could not be found for many model permutations. Other models generated illogical fit parameters consisting of $\sigma_o < 0$ or $n > 7$. Only one attempt generated plausible fit parameters: the natural logarithm-transformed 765 K isotherm for pre-production rolled GRCop-84 (Fig. 6). The stress exponent of the threshold stress model (5.3) is less than that of the power law model (7.6), as expected. Interestingly, this is the only data set among those collected that suggests the presence of a threshold at the lowest stress (Fig. 3). The threshold stress
model may apply to rolled GRCop-84, but it does not apply to extruded GRCop-84.

5.4. Rösler–Arzt model

The RA model was applied to both extruded and pre-production rolled GRCop-84 creep data using the Non-linear Regression function within *Mathematica*. All data within an isotherm were fit to a single equation using two different fit schemes. In both cases, temperature-dependent shear modulus was again computed from the expression given by Frost and Ashby [21]. In the first method, the only fit parameter was dislocation density (\( \rho \)). The second method fit the model to both dislocation density (\( \rho \)) and the activation energy for vacancy diffusion (\( Q_c \)). Unfortunately, no scheme was devised to permit a simultaneous fit on all five isotherms. Therefore, each isotherm was treated as a separate fit, even though the number of data points for these isothermal fits is too small to yield good statistics.

This numerical method was first calibrated by applying it to GlidCop Al-15 creep data at 773 and 973 K from Broyles et al. [18]. The Orowan stress, \( \sigma_o \), interaction parameter, and detachment stress, \( \sigma_d \), were all computed based on choosing \( Q_c \), \( n_{app} \), \( T \), and a variety of other constants (including \( \lambda \), \( b \), \( \nu \), \( M \), etc.). Poisson’s ratio was adjusted to generate identical values of the Orowan stress listed in their Table 4 [18], yielding \( \nu = 0.3 \). Appropriately natural logarithm-transformed isothermal pairs of data (applied stress and observed strain rate) were then fit to determine dislocation density. In comparing results to those reported by Broyles et al., interaction parameter magnitudes are duplicated (except for one), detachment stresses are within 6%, and dislocation densities generated ranged from 9 to 2 orders of magnitude lower than those reported by Broyles et al. [18]. While these results are not numerically identical, they indicate that the basic methodologies employed were consistent with the methods used by Broyles et al. [18].

The RA model was then applied to pre-production rolled and extruded GRCop-84 in the same manner as described above. Where possible, constants (including \( \nu \)) used in the GlidCop Al-15 analysis were used in this analysis. The volume fraction of Cr\(_7\)Nb is 14.1% [1], but this number includes the nearly micron sized particles which are too large to interact with dislocations and, therefore, cannot contribute to the dislocation detachment mechanism. The volume fraction of small particles was estimated to be 5% (0.05) and particle diameters of 10, 30, and 50 nm were considered. The smallest particles known to exist in GRCop-84 have a diameter of 30–50 nm [2–4]. Fitting only dislocation density, regardless of the choice of \( Q_c \) for various diffusion mechanisms in copper (117 kJ/mol \( \leq Q_c \leq 200 \) kJ/mol), the interaction parameter was greater than 0.94 for particle diameters of 50 and 30 nm. Values of \( k > 0.94 \) means the RA model does not apply because there is no detachment control. Even if there were particles with a diameter as small as 10 nm, the interaction parameter would be weak (0.921 \( \leq k \leq 0.934 \)). Fitting both \( Q_c \) and \( \rho \) either generated \( k > 0.94 \) or unreasonable dislocation densities (\( \rho > 10^{20} \) m\(^{-2} \)). In summary, the RA model does not apply to GRCop-84 unless there exist very small particles, but these have not been previously observed.

5.5. Hyperbolic sine interpolation

Closer inspection of Fig. 5 suggests a transition at \( \sigma/\sigma_0 \) of approximately 10\(^{-3} \), with the limited data below this value possibly obeying a power law relationship and the data above this value exhibiting some stronger functionality. The curvature of these lower temperature and corresponding higher stress creep data suggest the presence of power law breakdown. Failure of the power law model at high stresses is
well known, especially for \( \sigma/G \geq 10^{-3} \) \cite{21,26}, although a detailed microstructural understanding of power law breakdown does not exist. However, it can be phenomenologically described by a hyperbolic sine relationship of the form:

\[
\frac{\varepsilon}{\dot{\tau}} = A \sinh \left( \frac{\sigma}{\alpha \dot{\tau} G} \right)^n
\]  

(7)

where \( D \) refers to diffusivity, \( A \) is a model constant, \( \alpha \) and \( n \) are both positive, dimensionless constants, and the other terms have their usual meaning. The constants \( A \) and \( n \) cannot be correlated to constants of the same designation from other models (e.g., power law model or threshold stress model). The principal utility of Eq. (7) is to facilitate interpolation to conditions for which no creep data exist. This model can, of course, also be used to extrapolate outside of the region where data exist, provided that no change in operative creep mechanism occurs.

Eq. (7) was applied to extruded GRCop-84 creep data using the Nonlinear Regression function of Mathematica. Since diffusivity is included in this equation, all isotherms were simultaneously fit. Data were provided either as pairs of \( \sigma/G \) and \( \dot{\varepsilon} \) or \( \dot{\varepsilon} = \exp(-Q/RT) \) or \( \sigma/G, \dot{\varepsilon} \) and \( T \), where \( D_{eff} \) is the effective diffusivity \( (D_{eff} = D_0 + \beta(\sigma/G)^2D_L) \), where \( D_0 \) is the lattice diffusion coefficient, \( D_L \) is the self-diffusion coefficient in the dislocation core, and \( \beta \) is a constant factor of about 10) and shear modulus was computed as described above for the threshold stress model. The hyperbolic sine model was evaluated as untransformed data and as natural logarithm-transformed data. In the latter case, the natural logarithm was taken of both strain rate and of Eq. (7). A variety of fittable parameters were explored, including \( A, Q \) (simple Arrhenius expression) or \( D_{eff}, n, \) and \( A \) fittable or \( \alpha \) equal to unity.

A satisfactory solution to this model was obtained from a natural logarithm-transformed data set. For this fit, the previously calculated average activation energy (286.6 kJ/mol) was used, \( A, \alpha, \) and \( n \) were all fittable parameters. Fig. 7 shows this fit for extruded GRCop-84 creep data. The six lowest \( \sigma/G \) data points reveal the onset of power law creep with a stress exponent of 4.9, consistent with dislocation climb controlled creep. These include the four data points at 1068 K that clearly showed a power law exponent of 4.7 (Fig. 3). Additional data at higher temperatures and/or lower stresses are needed to substantiate this idea of climb-controlled power law creep below \( \sigma/G \approx 10^{-3} \).

As discussed earlier, the master plot shown in Fig. 5 was developed using the activation energy computed from the power law model. The previous paragraph suggests that the power law model does not apply for most of the extruded GRCop-84 data set. This apparent disparity can be explained in the following manner. Each isotherm covers less than four orders of magnitude in strain rate, which is not a large enough range to observe the curvature caused by power law breakdown (in \( \dot{\varepsilon} \) versus \( \sigma \) space). The power law model generates acceptable statistics due to the small region being modeled at any one temperature. When the data are normalized to create the master plot, however, the normalized strain rates span nine orders of magnitude and the curvature associated with power law breakdown is apparent. Most of the GRCop-84 creep data lie in the power law breakdown region because this material is extremely strong and in order to obtain reasonable creep rates, the applied stresses must be large. Furthermore, since most of the creep data collected lie in the power law breakdown region, the threshold stress and RA models are not expected to apply.

5.6. Monkman-Grant relationship

The product of minimum creep rate and rupture time in homologous units is a constant. This observation was first made by Monkman and Grant in 1956 and is represented by the relationship, \( \dot{\varepsilon} = k_{\varepsilon} \times \exp(-Q/RT) \) \cite{27}. This relationship holds for most structural materials and indicates that fracture is controlled by creep mechanisms. In particular, the effects of stress and temperature on rupture time are identical to those for creep. Fig. 8 shows that extruded and pre-production rolled GRCop-84 follow the Monkman-Grant relation quite well. Since both material condition types are represented equally well (within some
scattered), the fracture mechanism is the same for both rolled and extruded material.

6. Correlation of creep properties with microstructure

GRCop-84 exhibits tensile ductility typically in excess of 10% true strain at creep temperatures and in excess of 13% engineering strain at room temperature. Fracture surface and surface damage features of ten specimens were systematically examined to correlate the ductility and creep properties noted above.

These observations show that creep-tested, extruded GRCop-84 fails by void growth and coalescence coupled with extensive creep crack growth and interlinkage. Deformation and damage are spread across the width of the specimen and significant surface cracking extends well away from the fracture surface. For large dimples, dimple diameter decreases with an increase in applied stress under isothermal conditions or with an increase in temperature under isostress conditions.

Some extruded GRCop-84 creep specimens show a bimodal dimple size on the fracture surface where a single large dimple (average size approximately 5–15 μm) is surrounded by much smaller dimples (average size approximately 0.5–1 μm) located on the ridges dividing large dimples (Fig. 9). While the data set is small, there does not appear to be a correlation of bimodalism with stress, temperature, strain rate, or condition (rolled versus extruded). Pre-production rolled GRCop-84 exhibits small dimples, but they are not distributed around the margins of large dimples.

At creep temperatures, the fracture surface of pre-production rolled GRCop-84 exhibits a layered appearance at magnifications less than approximately 3500×. These layers are parallel to the rolling plane and can be found most prominently near the specimen faces (Fig. 10). Layer separation appears to be caused by the formation and growth of dimples transverse to the loading axis. Some layer structures have ribs normal to the layer, suggesting that ribs are remnant walls between two dimples. Dimples on the fracture surface are rare and large dimples are absent in pre-production rolled GRCop-84 as compared to extruded material.

Large particle-free, very smooth areas were frequently observed on the fracture surfaces in creep tested GRCop-84. These areas are presumed to be sheared matrix material that has relaxed due to diffusion and surface tension effects. As
noted earlier, creep experiments take several hours to cool to room temperature, thereby providing time and temperature for such diffusion to occur even on the final ligament remaining prior to fracture. These smooth particle-free areas are present in both material types, but at a reduced area fraction in pre-production rolled material. Some fracture surface regions in extruded material are nearly particle free and show evidence of slip in the form of faint contour lines.

Extensive cracking and crack bridging were observed on the faces of all creep-tested specimens. One pre-production rolled specimen revealed bridging structures in cracks 2 mm from the fracture surface. The cross sectional shape of these structures cannot be determined from the views taken, but
the photomicrographs suggest that they have a complex cross section (Fig. 11, e.g., polygonal or “+” shape). It is not known, however, if crack bridging occurs other than at the specimen surface.

7. Concluding remarks

The constant, true applied stress tensile creep behavior of extruded GRCop-84 has been measured from 764 to 1068 K in vacuum. This alloy exhibits pure metal type creep curves with secondary creep occurring over a one to two percent true strain and true failure strains typically greater than 10%. Additionally, the Monkman–Grant relationship is obeyed for both pre-production rolled and extruded GRCop-84.

Stress exponents for extruded GRCop-84 range from 8.7 to 4.7, making them too large to directly apply to models of creep deformation in pure metals. A typical Arrhenius calculation of average creep activation energy using these stress exponents yields an average value of 287±14 kJ/mol. This value is larger than the activation energy for vacancy diffusion in pure copper of 197 kJ/mol, which is a typical result for a particle-strengthened metal. This power law analysis has limited utility since a master plot shows power law breakdown in extruded GRCop-84 for $\sigma/G_0 > 10^{-3}$.

The threshold stress model appears to fit pre-production rolled creep data at 765 K. Neither the threshold stress model nor the Rosler–Arzt model applies to extruded GRCop-84 creep data. These models would not be expected to apply for power law breakdown conditions. The successful model will likely need to describe strengthening by dislocations interacting with very small particles as well as a reduction in stress acting on those dislocations caused by a large number of nearly grain-sized Cr2Nb particles.

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