THERMAL ACTIVATED GRAIN BOUNDARY CREEP IN POLYCRYSTALLINE COPPER
(A Review)

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SUMMARY
Creep deformation in metals and alloys at intermediate temperatures and low stresses are attributed to power-law and diffusion mechanisms. Thermal activation parameters of steady state creep correlate with the macroscopic and microscopic variables, leading to inter-relationships between the apparent and true parameters. The observations are confirmed by tensile creep data of high purity polycrystalline copper, obtained by stress incremental and temperature differential methods at intermediate temperatures of 573, 673 and 773 K and low stresses of 7.08, 14.16, and 21.24 MPa, for creep rates ranging from $10^{-7}$ to $10^{-5}$ s$^{-1}$. The mean apparent (experimental) stress exponent, activation energy and activation volume of 2, 60 kJ/mol and 1.5 x $10^{-25}$ m$^3$, are lower than the true (model) values, suggesting that grain boundary phenomenon is rate controlling, and the difference in values can be associated with the limitations imposed by the internal stress to creep. The decrease of stress exponent with temperature increase indicates transition from power-law creep to Newtonian flow, or progressive weakening of the grain boundaries with enhanced relative movement of the grains. To a first approximation, independent deformation mechanisms or individual strain rates can be coupled to obtain the net strain rate. By superposition of the rate equations, the net strain rate is determined as the sum of three independent creep mechanisms of Cobble diffusion creep, grain boundary sliding and power-law creep. The stress exponent 2 relates to grain boundary sliding, while the lower activation energy with respect to the lattice self-diffusion energy signifies grain boundary movement. Therefore, the most plausible mechanism for the grain boundary diffusion coefficient being higher than the lattice diffusion coefficient in the creep test regime is grain boundary movement. LINEST multiple regression analysis of the creep data establishes excellent correlations between the strain rate, applied stress and temperature; and the best fit creep equation gives values of the regression parameters which are consistent with the empirical values.

Keywords: Creep deformation, Thermal activated parameters, Cobble diffusion creep, Grain boundary sliding, Power law creep, Apparent and True parameters

1. INTRODUCTION
At high temperatures, metals and alloys show rate dependence deformation, which is thermally activated time dependent process, and as the temperature increases, the stress required to produce plastic deformation decreases. Creep occurs through many interacting processes, but the common mechanisms are dislocation slip and climb, diffusion creep, and grain boundary sliding. As more than one mechanism can operate at the same time (in parallel), independently of each other, the total creep rate is the sum of individual rates; and the fastest mechanism will control the creep behaviour, while the slowest controls the deformation [1].

If two or more mechanisms of different activation energies contribute to the creep rate, the overall activation energy depends on the test temperature, and marginally on the mechanical stress through the activation volume [2]. Identification of creep mechanisms is necessary for prediction of creep in solids, and also for developing innovative creep resistance materials.

Mathematical analysis of thermal activated creep has been based on rate theory of independent creep mechanisms, differential stresses, temperature changes, and limitation imposed by internal stress of the material [1 - 13].

Creep deformation is characterized by time-dependent macroscopic variables of stress, temperature, strain and strain rate. When the stress and temperature are prescribed as independent variables, the strain rate is determined; but if temperature and strain rate are specified, the stress can be determined. The microscopic or state variables are specified by dislocation density, point defects, grain size, and lattice parameters (i.e. atomic volumes, bond energies, elastic moduli, diffusion constants, etc.).

The state variables control dislocation glide and vacancy diffusion, and hence the creep deformation; and should not be regarded as constant, except when the creep properties of different materials are being compared. For instance, no creep could occur at a threshold stress where the strain rate is zero, otherwise an eff-
effective stress would determine the deformation. An internal stress of zero threshold value for creep indicates that the microstructure remains constant with time and the creep rate is at steady state.

For low stress and intermediate temperature creep, the dominant deformation mechanism is grain boundary movement. The macroscopic parameters imposed during creep are assumed to be different from the parameters associated with microscopic diffusion creep, since in general, the shear strain rate is not the same as the measured tensile strain rate, due to possible existence of microscopic heterogeneities [3].

In this review, stress exponent, activation energy and activation volume data of steady state creep of polycrystalline copper are examined in relation to the thermal activated parameters associated with the macroscopic and microscopic variables linked with grain boundary movement. In addition, the relationships between the creep rate, applied stress, temperature, and internal stress due to microstructure and defects of the material are considered.

2. CREEP DEFORMATION

The deformation processes which produce permanent plastic deformation in metals and alloys have been developed into deformation-mechanism maps by Frost and Ashby [13]. In the absence of any defects, plastic deformation is induced when the applied stress exceeds the threshold strength of 0.4μ, where μ is the shear modulus. Dislocations in metals enable deformation to occur at stresses well below the theoretical shear strength. Through thermal activated processes and depending on the spacing and strength of the microstructural obstacles, dislocation glide is induced leading to creep.

Diffusional creep of atoms along grain boundaries (Coble creep) is important at moderate stresses and temperatures, particularly in fine-grained metals. Nabarro-Herring diffusional creep via vacancy self-diffusion in the grain interior is significant at very high temperatures. At elevated temperatures and stresses, dislocation creep (power-law creep) occurs by dislocation climb and glide depending on the vacancy self-diffusion coefficient. Other forms of creep deformation are Harper-Dorn creep and diffusional creep by pipe diffusion.

The deformation mechanism map for Cu of 100 μm grain size is shown in Fig. 1, and typical values of creep parameters for Cu are listed in Table 1 [13]. The map shows three principal creep fields of low temperature plasticity, power-law creep and diffusional flow. Diffusion controlled creep is slower, as reflected by lower creep rates. The primary references for high temperature creep of Cu are Feltham and Meakin [14] and Barrett and Sherby [15]. Like other face-centred cubic metals, Cu has extremely low lattice resistance (< 10⁻⁷μ), and at about 0.3 T_M, starts to creep, where T_M is the melting temperature (K).

In Fig. 1, the creep field is subdivided into a region of low-temperature, core diffusion controlled creep in which the stress exponent is n + 2, and a region of high temperature lattice diffusion controlled creep in which the stress exponent is n. The power law breakdown occurs at σ_0/μ ≈ 10⁻³, where σ_0 is the steady state creep stress. Diffusional flow appears at high temperatures and low stresses, and consists of boundary diffusion and lattice diffusion controlled creep. The diffusional flow is limited by lattice diffusion (Nabarro-Herring creep) and grain boundary diffusion (Coble creep). The power law creep is limited by glide processes, lattice diffusion controlled climb (high temperature creep), and core diffusion controlled climb (low temperature creep).
\[
\dot{\varepsilon} = \sum a_i t^{n_i - 1}
\]
and for a single term (i = 1) of the equation describing a time dependence steady state creep,
\[
\dot{\varepsilon} = a t^{m - 1}
\]

There is a correlation between steady state creep parameters and temperature dependent Gibbs free energy [5]. If the free energy for activation or Gibbs free energy (\(\Delta G\)) is the thermodynamic state function, the stress \(\sigma\) and absolute temperature \(T\) are independent state variables, and \(\Delta G = \Delta G(\sigma, T)\) is related to the activation enthalpy \(\Delta H\), through the isothermal Gibbs-Helmholtz relationship,
\[
\Delta G = \Delta H - T \Delta S
\]
where the activation entropy is defined as
\[
\Delta S = \left( \frac{\partial Q}{\partial T} \right)_\sigma
\]
and \(Q\) is activation energy for the creep process. \(\Delta G(\sigma, T)\) is dependent on the internal stress and distribution of obstacles. Combining eqns. (4) and (5),
\[
\Delta G = \Delta H + T \left( \frac{\partial Q}{\partial \sigma} \right)_\sigma
\]

For deformation process where change in enthalpy equals the activation energy, \(\Delta G = 0\), and \(\Delta H = \Delta Q\)
\[
\text{Equation (6) provides functional forms for the free energy determined from experimental data.}
\]

From eqn. (1), \(\ln \dot{\varepsilon} = f(\sigma, T)\), and the partial derivatives, \(\frac{\partial \ln \dot{\varepsilon}}{\partial \sigma}\), \(\frac{\partial \ln \dot{\varepsilon}}{\partial T}\), \(\frac{\partial \ln \dot{\varepsilon}}{\partial \sigma}\), \(\frac{\partial \ln \dot{\varepsilon}}{\partial T}\), and \(\frac{\partial \ln \sigma}{\partial \varepsilon}\), \(\frac{\partial \ln \varepsilon}{\partial \sigma}\), \(\frac{\partial \ln \varepsilon}{\partial T}\), and \(\frac{\partial \ln \varepsilon}{\partial \sigma}\) are inter-related and expressed in the general form [16],
\[
\left[ \begin{array}{c}
\frac{\partial \ln \dot{\varepsilon}}{\partial \sigma} \\
\frac{\partial \ln \dot{\varepsilon}}{\partial T} \\
\frac{\partial \ln \dot{\varepsilon}}{\partial \sigma}
\end{array} \right] = -1
\]
In explicit form, eqn. (2) becomes
\[
\Delta H = - R \left[ \frac{\partial \ln \dot{\varepsilon}}{\partial (1/T)} \right]_\sigma
\]

For diffusion controlled creep, where the activation enthalpy equals the apparent activation energy \(Q_a\), the stress is related to the strain rate through a power law (\(\dot{\varepsilon} \propto \sigma^{n_a}\)), while the strain rate and temperature are related exponentially (\(\dot{\varepsilon} \propto \exp \left( \frac{Q_a}{RT} \right) \)). Under the two conditions, eqn. (1) is transformed into the form,
\[
\dot{\varepsilon} = A \sigma^{n_a} \exp \left( - \frac{Q_a}{RT} \right)
\]
for analyzing experimental creep data, where \(A\) is a constant, \(n_a\) is apparent or measured stress exponent and \(Q_a\) is apparent activation energy. The apparent stress exponent depends on the testing conditions, but is typically ~2 for grain boundary sliding, ~1 for Newtonian viscous flow, and >1 for non-Newtonian viscous behavi-
our. From eqn. (8), two activation enthalpies can be defined mathematically for conditions of constant stress or constant strain rate, but the creep deformation process has a unique activation energy, which is independent of the method of determination, either via eqn. (2) or (9) [5].

If for the testing conditions $\sigma/\mu \sim 10^{-2}$ and $T \sim 0.5T_M$, $Q_i$ is equal to the activation energy for lattice self-diffusion ($Q_i$), then dislocation climb is rate controlling; but if $Q_a \approx 0.6 Q_i$ for $\sigma/\mu < 10^{-4}$ and $T \leq 0.5T_M$, then grain boundary sliding is rate controlling.

3.2. Independent Creep Mechanisms

For creep deformation where independent thermally activated mechanisms are operating, the observed creep rate, $\dot{\varepsilon}_{obs}$, is the sum of the strain rates due to each mechanism [17];

$$\dot{\varepsilon}_{obs} = \sum_{i=1}^{n} \dot{\varepsilon}_i = \sum_{i=1}^{n} A_i \varepsilon_i^{n_i} \exp \left( \frac{-Q_i}{RT} \right)$$  (11)

where the creep parameters $A_i$, $Q_i$, and $n_i$ are respectively, the structural parameter, activation energy and stress exponent for each mechanism. Depending on the exact value of the parameters, each term in eqn. (11) might dominate all others over some range of $\sigma$ and $T$ values.

If the strain rates are measured over range of stresses and temperatures where only one creep mechanism dominated, then the creep parameters can be determined for that particular mechanism. While the apparent activation energy ($Q_a$) and apparent stress exponent ($n_a$) are determined separately from temperature differential and stress change tests, all the three parameters ($A$, $Q_a$, $n_a$) can be evaluated by multiple linear regression analysis of the creep data.

3.3. Weighted Parameters of Independent Mechanisms

An unequivocal determination of the dominant creep mechanism depends on the material properties ($A_i$, $Q_i$, and $n_i$) and experimental conditions ($\sigma$, $T$ and $\dot{\varepsilon}$). The observed stress exponent and apparent activation energy can be evaluated respectively by the equations [17],

$$\dot{\varepsilon}_{obs} = \sum_{i=1}^{n} A_i \varepsilon_i^{n_i} \exp \left( \frac{-Q_i}{RT} \right)$$  (12)

$$Q_{obs} = \frac{\sum_{i=1}^{n} Q_i \dot{\varepsilon}_i}{\sum_{i=1}^{n} \dot{\varepsilon}_i} \sum_{i=1}^{n} Q_i A_i \varepsilon_i^{n_i} \exp \left( \frac{-Q_i}{RT} \right)$$  (13)

In either eqn. (12) or eqn. (13), the calculated value is a weighted sum of either the stress exponent or the apparent activation energy, where the weighting parameter in each case is the strain rate for each independent creep mechanism. As the strain rate for the i-th mechanism increases with increasing temperature and stress, the weightings given the i-th observed strain rate, stress exponent or activation energy increase to a limiting value of $A_i \sigma^{n_i}$ [17].

4. ANALYSIS OF THERMAL ACTIVATED PARAMETERS

4.1. Apparent and True Activation Parameters

An experimental measured creep parameter is termed “apparent”, whereas “true” parameter is calculated from the constitutive equation representing the microscopic variables of the deformation. That is, apparent values $n_a$, $Q_a$ and $\Delta V_a$ are empirically determined, while the true values $n_i$, $Q_i$, and $\Delta V_i$ are deduced from model equations based on the limitations imposed by the lattice, microstructure and deformation regimes.

The activation parameters used to identify possible creep mechanisms are fundamental for providing innovative methods for developing creep resistance materials. Many different mechanisms might interact to initiate the activation process, but cannot be separated based only on experimental data, hence the development of creep deformation maps [13]. An apparent parameter may be different from the true parameter because of possible existence of threshold internal stress. In the absence of internal stress, the true stress exponent is equivalent to the apparent value.

In eqn. (10), the term $A$ is not strictly independent of temperature, and the strain rate $\dot{\varepsilon}$ of a diffusion-controlled process has been represented by Dorn’s equation [10],

$$\dot{\varepsilon}_i = A \left( \frac{\partial \ln \dot{\varepsilon}_i}{\partial \ln \sigma_i} \right) \frac{\dot{\varepsilon}_i}{\sigma_i} \left( \frac{\sigma_i}{\mu} \right)^{n_i}$$  (14)

where $A_i$ and $z_i$ are dimensionless parameters (which depend on the specific assumptions inherent in the deformation mechanisms), $\sigma$ is grain size and $b$ is magnitude of Burgers vector. The inverse grain size exponent $z_i$ is given by

$$z_i = \frac{\partial \ln \dot{\varepsilon}_i}{\partial \ln \sigma_i}$$  (15)

and the stress exponent is defined by

$$n_i = \frac{\partial \ln \dot{\varepsilon}_i}{\partial \ln \sigma_i}$$  (16)

The diffusion coefficient $D$ of the rate-controlling mechanism is defined by,

$$D = D_o \exp \left( \frac{-Q_i}{RT} \right)$$  (17)

where $D_o$ is a frequency factor and $Q$ is enthal-
py of the diffusion process. Equation (14) has advantage over eqn. (10), for placing the stress in a normalized form, $\sigma/\mu$. Dependence of the activation enthalpy on internal stress limits the deformation to give the true activation energy $Q_i$. From eqn. (17), the true activation energy for diffusion controlled grain boundary creep is defined as,

$$Q_i = -R \left[ \frac{\partial \ln \frac{\sigma}{\mu}}{\partial (1/T)} \right]_{d, \sigma}$$ (18)

Equation (10) has been used extensively to determine the creep parameters ($A$, $Q_i$, and $n_i$) for a number of materials under specific experimen\-tal conditions at which a single creep mechanism controlled the net steady state creep rate, and $Q_i > Q_e$ for mechanisms where $n = 1$, but $Q_i > Q_e$ at high temperatures and when $n \geq 1.5$.

The experimental values of $n_i$, $Q_i$ and $\Delta V_s$ determine the rate controlling mechanisms for the thermal activated deformation [1, 3]. Lattice mechanisms have non-zero value of $z$; but for grain boundary sliding with plastic flow, $z = 1$. The value of $Q_i$ corresponds to the self-diffusion energy for diffusion controlled mechanism in the grains, within the grain boundaries, or by dislocation pipe diffusion. The parameter $n_i$ is significant in identifying the substructure mechanism, whether by dislocation glide or grain boundary sliding. For crystalline materials, $n = 1$ is associated with creep controlled by diffusion, but dislocation models predict $n \geq 3$, while grain boundary sliding mechanism is characterized by $n = 2$.

4.2. Internal Stress and Stress Exponent

A variable for describing overall effect of microstructural evolution on creep response is the internal stress $\sigma_i$, which results from the presence of obstacles, long range barriers, strain hardening; and opposes dislocation motion. At steady state condition, strain hardening and recovery balances, and $\sigma_i$ is limited to a fraction of the applied stress $\sigma_a$. The internal stress level at which the microscopic creep strain is zero is the threshold stress $\sigma_{th}$.

In the absence of internal stress $\sigma_i$, the power-law relations for stress dependence is,

$$\dot{\varepsilon} = A_0 \sigma_a^n$$ (19)

where $\sigma_a$ is the applied stress and the apparent stress exponent is defined as,

$$n_a = \left( \frac{\partial \ln \frac{\sigma_a}{\mu}}{\partial \ln \frac{\sigma_a}{\mu}} \right)_T$$ (20)

In the presence of internal stress, the difference between applied and internal stress is the effective stress, $\sigma_e = \sigma_a - \sigma_i$, and the steady state power-law creep eqn. (19) becomes,

$$\dot{\varepsilon} = A(\sigma_a - \sigma_i)^{n_1}$$ or $$\dot{\varepsilon} = \left( \frac{\partial \ln \frac{\sigma_a}{\mu}}{\partial \ln (\sigma_a - \sigma_i)} \right)_T$$ (21)

where $n_e$ is the true stress exponent,

$$n_e = \left( \frac{\partial \ln \frac{\sigma_a}{\mu}}{\partial \ln (\sigma_a - \sigma_i)} \right)_T$$ (22)

From plots of eqn. (21), the threshold stress $\sigma_{th}$ below which no creep deformation would occur at a given T is obtained when $\dot{\varepsilon} = 0$.

Load transfer compensate to internal resistance, slows down creep rate and improves creep resistance. If the load transfer only occurs when the external or applied stress exceeds threshold stress, thereby causing creep deformation, then the effective stress could be expressed by [18],

$$\sigma_e = (1-\alpha)(\sigma_a - \sigma_{th})$$ (23)

where $\alpha$ is a load transfer factor, with value ranging from 0 to 1. When $\alpha = 0$, there is no load transfer; while $\alpha = 1$ indicates that the loads are transferred to the testing material.

4.3. Activation Energy

Activation energy $Q_a$ is designated apparent for creep because the value is empirical determined, and is given by

$$Q_a = -R \left[ \frac{\partial \ln \frac{\sigma}{\mu}}{\partial (1/T)} \right]_{d, \sigma}$$ (24)

The true activation energy for creep is a function of the apparent activation energy, temperature, material structure and elastic modulus of the deforming solid; related by the expression [5],

$$Q_i = Q_a + RT \left[ 1 + \left( \frac{n-1}{n} \right) \frac{\partial \mu}{\partial T} \right]_{d, \sigma}$$ (25)

From eqn. (25), $Q_i$ is determined using modulus compensated stress $\sigma/\mu$, at T. Recognizing that the effect of T on $\mu$ of material yields $Q_i$, eqn. (25) can be modified where $Q_i$ is corrected by variation in $\mu$ with T as [19],

$$Q_i = -R \left[ \frac{\partial \ln \frac{\sigma}{\mu}}{\partial (1/T)} \right]_{d, \sigma} - nR \left[ \frac{\partial \ln \mu}{\partial T} \right]_{d, \sigma}$$ (26)

where $Q_i$ is contribution of change in modulus as a function of T. If the effect of T on $\mu$ is small, $Q_i$ approaches zero, and $Q_i = Q_a$.

For polycrystalline Cu of $d = 100 \mu$, $\mu = 4.75 \times 10^4$ - 17T (MPa), $\left( \frac{\partial \ln \sigma}{\partial T} \right)_{d, \sigma}$ is negative [13], and eqn. (25) gives $Q_i < Q_a$ for $n = 1$, or $Q_i < Q_a$ for $n > 1$, depending on the testing T. The values of $Q_i$ and $Q_a$ are generally different because of the term $\mu^{1/(n-1)}T$. The true activation energy is obtained from the slope of a plot of $\ln \dot{\varepsilon} \mu^{1/T}$ versus 1/T [5]. The apparent activ-
tion energy can be lower than the true activation energy, depending on the stress [20].

The true activation energy \( Q_t \) also accounts for both glide and recovery processes in steady state creep and shows dependence on internal stress \( \sigma_i \). To account for linear dependence of \( \ln \dot{\varepsilon} \) on \( \ln \sigma \) at \( T \), there must be a linear dependence of \( Q_t \) on \( \sigma \), hence [21],
\[
Q_t = Q_a - \Delta V \sigma_e
\]
(27)
which indicates that \( Q_t \) is determined by competition of effective stress \( \sigma_e \) and activation volume \( \Delta V \).

### 4.4. Activation Volume

The apparent activation volume \( \Delta V_a \) related to the mechanical work involved in the creep process decreases with increasing total strain, and is expressed as change of rate of deformation as a function of change of stress at constant temperature [22],
\[
\Delta V_a = kT \left( \frac{\partial \ln \dot{\varepsilon} \sigma_i}{\partial \sigma} \right)_T = n_a \frac{kT}{\sigma}
\]
(28)
Also if the stress dependence of entropy term in the rate eqn. (1) is negligible, \( \Delta V_a \) for creep is estimated as [23],
\[
V_a = -2 \left( \frac{\partial K_{\Delta \sigma}}{\partial \sigma} \right)_T
\]
(29)
If \( \tau \) is shear stress, and assuming \( \sigma \approx 2\tau \), then \( V_a = 245b^2 \), where \( b \) is Burgers vector [23].

### 5. DETERMINATION OF ACTIVATION PARAMETERS

#### 5.1. Strain Change at constant Temperature

If only external stress \( \sigma_e \) is applied at a specified \( T \) (absence of internal stress), the apparent stress exponent is defined as,
\[
n_a = \left( \frac{\partial \ln \dot{\varepsilon} \sigma_i}{\partial \sigma} \right)_T
\]
(20)
In the presence of internal stress at constant temperature, the true stress exponent is determined as,
\[
n_t = \left( \frac{\partial \ln \dot{\varepsilon} \sigma_i}{\partial \ln (\sigma_{\sigma} \sigma_i)} \right)_T
\]
(22)

#### 5.2. Change in Temperature at constant Stress

If \( T \) is changed rapidly during a test at constant \( \sigma \), the apparent activation energy based on eqn. (24) is given as [11, 12],
\[
Q_a \approx \frac{R \ln (\dot{\varepsilon}_1 / \dot{\varepsilon}_2)}{(T_2 - T_1) / T_1 T_2}
\]
(30)
where \( \dot{\varepsilon}_1 \) and \( \dot{\varepsilon}_2 \) are the instantaneous creep rates immediately preceding and following a small increase in temperature from \( T_1 \) to \( T_2 \). If the apparent stress exponent \( n_a \) is known for the conditions of the test, the true activation energy, \( Q_t \) can be determined from the relationship [5],
\[
Q_t = R \left[ \frac{\partial \ln \dot{\varepsilon} \mu_{\mu_a+1}^1}{\partial (1/T)} \right]_{\sigma} = \frac{R (\dot{\varepsilon}_2 \mu_{\mu_a+1}^2 - \dot{\varepsilon}_1 \mu_{\mu_a+1}^1)}{(T_2 - T_1) / T_1 T_2}
\]
(31)
where \( \mu_1 \) and \( \mu_2 \) are the values of the shear modulus at \( T_1 \) and \( T_2 \), respectively.

#### 5.3. Change in Strain rate at constant Stress

Under conditions of constant stress, \( Q_s \) and \( Q_t \) are given by eqns. (30) and (31), respectively. A plot of \( \log \dot{\varepsilon} \) versus \( 1/T \) has a slope of \( Q_s / 2.3R \), whereas a plot of \( \log \dot{\varepsilon} \mu_{\mu_a+1}^1 \) versus \( 1/T \) has a slope of \(-Q_t / 2.3R\).

#### 5.4. Change in Stress at constant Strain rate

Applying eqns. (9) and (24) under conditions of constant strain rate, \( Q_s \) and \( Q_t \) are determined from the relationships,
\[
Q_s = nR \left[ \frac{\partial \ln \dot{\varepsilon} \sigma_i}{\partial (1/T)} \right]_{\sigma, k}
\]
(32)
and
\[
Q_t = R \left[ \frac{\partial \ln (\sigma^\mu_{\mu_a+1}^1 \mu_{\mu_a+1}^2 - \dot{\varepsilon} \mu_{\mu_a+1}^1)}{\partial (1/T)} \right]_{\sigma, k}
\]
(33)
A plot of \( \log \sigma \) versus \( 1/T \) has slope of \( Q_s / 2.3nR \), whereas a plot of \( \log \sigma^\mu_{\mu_a+1}^1 \) versus \( 1/T \) has slope of \( Q_t / 2.3R \).

#### 5.5. Stress derivative of Activation Energy

By eqn. (29), the stress derivative of activation energy at constant \( T \) is the activation volume. Neglecting hydrostatic pressure, the apparent activation volume is given by,
\[
\Delta V_a = \frac{n_a kT}{\sigma_a}
\]
(34)
whereas the true activation volume is,
\[
\Delta V_t = \frac{n_t kT}{\tau}
\]
(35)
Experimentally determined activation volume is related to the true activation volume by the relationship [24],
\[
\Delta V_a = \alpha \Delta V_t
\]
(36)
where \( \alpha = 0.67 \) at low temperatures and \( \alpha = 0.5 - 0.6 \) at high temperatures.

The corresponding apparent activation area \( \Delta A_a \) and true activation area \( \Delta A_t \) are defined respectively by
\[
\Delta A_a = \frac{kT}{b} \left( \frac{\partial \ln \sigma_i}{\partial \sigma} \right)_T = \frac{\Delta V_a}{b} = \frac{n_a kT}{b} = -1 \frac{\partial \Delta A_a}{\partial \sigma} \left|_T \right.
\]
(37)
\[
\Delta A_t = \frac{n_t kT}{b} = -1 \frac{\partial \Delta A_t}{\partial \sigma} \left|_T \right.
\]
(38)
6. ANALYSIS OF CREEP DATA

The rate controlling mechanism(s), relationship between apparent and true parameters (or macrostructure and micro-structure variables), and coupling of power-law creep and diffusional flow variables are examined by applying the fundamental theories of creep to the deformation data of polycrystalline Cu.

6.1. Experiments

Hard-drawn polycrystalline copper wires of 99.9 % purity (with 0.03 % bismuth and lead as impurities, excluding oxygen and silver), d ~ 100 μm and nominal diameter of 2.10 mm were creep tested. A dead-loading Denison creep machine, incorporating high temperature heating equipment and linear variable differential transducer, was used to conduct the axial tensile creep tests at normal atmosphere, T of 573, 673 and 773 K (0.42 - 0.57 Tm), and σ of 7.08, 14.16 to 21.24 MPa (σ/μ < 10⁻⁴). Thermal equilibrium of the apparatus was attained 30 minutes after switching on, and the test chamber was maintained at ± 1 K. The creep strain rates ranged from 10⁻⁷ s⁻¹ to 10⁻⁵ s⁻¹.

The experimental data were analysed and the creep stress exponent, activation energy and activation volume determined and interpreted in line with inter-dependence of power-law and diffusion flow as creep mechanisms.

To minimize experimental scatter between specimens of the same grain size, stress incremental technique was adopted, and creep rates were measured at different stresses on one specimen under isothermal conditions. Upon application of the first stress σ₁, primary followed by steady-state creep were obtained. Following a period well into steady-state, the stress was increased from the initial σ₁ to a new level σ₂, resulting in another increment of primary creep which settled into steady-state, and later, the stress was increased to the final stress σ₃. Each specimen was tested at a single temperature to obtain the creep curves shown in Fig. 2.

Temperature differential tests were also performed to determine the apparent activation energy, when the stress was held constant, while the temperature was changed abruptly from T₁ to T₂, causing a change in strain rate.

6.2. Apparent Stress Exponent

The dependence of steady-state creep rate on applied stress (in the absence of internal stress) at the test temperatures is shown in Fig. 3, as logarithmic plots of \( \dot{\varepsilon} \) versus σ. For the stress differential creep tests, at constant T, a change from an initial stress σ₁ to final stress σ₂ caused a change in strain rate from \( \dot{\varepsilon}_1 \) to \( \dot{\varepsilon}_2 \). At steady state conditions, the crystal structure remains constant to first order after the stress change, and the apparent stress exponent is:

\[
\eta_a = \frac{d \log \dot{\varepsilon}}{d \log \sigma} = \frac{\log(k_2/k_1)}{\log(\sigma_2/\sigma_1)}
\]

If a single creep mechanism is dominant, eqn. (10) becomes a valid approximation to eqn. (1).
The values of apparent stress exponent \( n_a \), 1.9, 2.0 and 2.2 (at \( T = 773, 673 \) and 573 K respectively), with a mean value of \( \sim 2.0 \), are indicative of creep by grain boundary movement. The gradual decrease in the value of \( n_a \) as the temperature increased is interpreted as a progressive weakening of the grain boundaries and transition from power law creep to Newtonian flow. The observation suggests that an increase in temperature enhanced relative movement of the grains.

6.3. Threshold Stress and True Stress Exponent

The existence of threshold stress \( \sigma_{th} \) is verified and determined by plotting isothermal lines of \( \dot{\varepsilon}^{1/n_a} \) versus \( \sigma_a \) in a bilinear plot as shown in Fig. 4, using values of \( n_a = 1.9, 2.0, \) and 2.2 at \( T = 773, 673, \) and 573 K. Extrapolating the plots to \( \dot{\varepsilon}^{1/n_a} = 0 \) for the test temperatures, gives the intercept on the stress axis or the threshold stress \( \sigma_{th} \) to be zero value. The observation implies that hard-drawn polycrystalline Cu wires of 99.9 % purity does not exhibit threshold stress at which the specimens would not creep. Also the true stress exponent is the same as the apparent stress exponent, and the power law eqn. (19) is valid and eqn. (20) is applicable to creep deformation.

6.4. Apparent Activation Energy

Assuming the crystal structure remains constant immediately after temperature change from \( T_1 \) to \( T_2 \) causing a change in strain rate from \( \dot{\varepsilon}_1 \) to \( \dot{\varepsilon}_2 \), the apparent activation energy is determined by;

\[
Q_a = -2.3R \frac{\text{dlog}(\dot{\varepsilon})}{\text{d}(1/T)} = 2.3R \frac{T_1 T_2}{T_2 - T_1} \log \left( \frac{\dot{\varepsilon}_2}{\dot{\varepsilon}_1} \right)
\]

(39)

The value of \( Q_a \) obtained by plotting \( \log \dot{\varepsilon} \) against \( 1/T \) in Fig. 5 ranges from 59.0 to 61.0 kJ/mol (mean value of 60.0 kJ/mol), but decreases with increasing stress.

6.5. True Activation Energy

The true activation energy \( Q_t \) obtained from plots of \( \log(\dot{\varepsilon}^{1/T_1}) \) versus \( 1/T \) in Fig. 6, has value ranging from 61.0 to 63 kJ/mol, mean value of 62.0 kJ/mol, and decreasing with increasing stress.
Fig. 7. Activation energy dependence on applied stress

As shown in Fig. 7, the true activation energy is higher than the apparent activation energy; an indication that the rate-controlling mechanism is a grain boundary phenomenon, involving a sequential process of Newtonian flow and power law creep.

There is a decreasing functional relationship between stress and activation energy, and extrapolating the relationships to zero-stress, yields Q_s = 62.5 kJ/mol and Q_a = 64.5 kJ/mol. The zero stress activation energy represents the maximum potential barrier to be overcome for creep deformation.

As a general rule, the activation energy for grain boundary diffusion is ~ 0.5 Q_l (lattice diffusion energy) [25]. If the vacancy migration enthalpy is 80 kJ/mol [26], the zero stress activation energy ranging between 62 and 64 kJ/mol is higher than half-value of migration enthalpy. The observation confirms that the creep deformation is controlled by Coble grain boundary movement (Newtonian flow) accommodated by grain boundary sliding, coupled with power-law creep in the grain interior.

6.6. Apparent Activation Volume

The apparent activation volume is determined from the relation,

$$\Delta V_a = kT \left( \frac{d \log \dot{\varepsilon}}{d\sigma} \right)$$  (28)

Figure 8 is a plot of \( \log \dot{\varepsilon} \) versus \( \sigma \), and tangents were fitted to the curves to evaluate the gradient, and hence the value of \( \Delta V_a \).

The value of \( \Delta V_a \) ranges from 1.36 x \( 10^{-27} \) m³ to 1.7 x \( 10^{-27} \) m³, increases slightly with temperature (Fig. 9), but shows significant decrease with stress increase (Fig. 10). The inverse relationship is interpreted as work hardening leading to decrease in volume along the grain boundary where deformation occurs; and hence, a decrease in grain boundary width, compensated for by elongation and subsequent strain of the grains (as illustrated in Fig. 11). From Figs. 9 and 10, the activation volume for creep is observed to be strongly dependent on stress, but shows slight dependence on temperature.

6.7. True activation volume

Using eqn. (36), \( \Delta V_a = 0.67 \Delta V_s \), and the true activation volume \( \Delta V_s \) ranges from 2.04 x \( 10^{-27} \) m³ to 2.57 x \( 10^{-27} \) m³. Assuming the shear stress on slip plane \( \tau = \sigma/1.7 \) MPa [13], the maximum true activation work \( \tau \Delta V \) of 1.78 x \( 10^{-20} \) J is of the same order of magnitude as the lowest energy barrier for the creep deformation.

6.8. LINEST Analysis of Creep Data

Assuming that a single creep mechanism is dominant for the overall creep deformation, a multiple linear least square minimization procedure can be used to determine the apparent parameters \( A \), \( Q_a \), and \( n \), by linearizing eqn. (10) to the form,

$$\log \dot{\varepsilon} = \log A - \frac{Q_a}{2.303R} \frac{1}{T} + n \log \sigma$$  (40)

As shown in Figs. 3 and 5, there are excellent correlations between plot of \( \log \dot{\varepsilon} \) and \( \log \sigma \), and plot of \( \log \dot{\varepsilon} \) and \( 1/T \). Therefore, the
Fig. 9. Variation of activation volume with temperature

Fig. 10. Dependence of activation on applied stress

The multiple regression eqn. (40) linking the dependent variable ln $\dot{\varepsilon}$ and set of independent variables of $1/T$ and log $\sigma$ is of the form,

$$\log \dot{\varepsilon} = m_1 \log \sigma + m_2 \left( \frac{1}{T} \right) + A_0$$  \hspace{1cm} (41)

where $m_1$, $m_2$ and $A_0$ are the regression coefficients, equivalent to the values of $n$, $-Q/2.3R$ and log $A$, for the best empirical equation fit to the experimental data, $\dot{\varepsilon} = f(T, \sigma)$.

By applying Microsoft Excel LINEST function [27], the regression coefficients and the statistics of standard deviation, coefficient of determination $r^2$, F-statistic and degree of freedom determined are shown in Table 2.

<table>
<thead>
<tr>
<th>Log $\sigma$ (MPa)</th>
<th>$1/T$ (K$^{-1}$)</th>
<th>Log $\dot{\varepsilon}$ (s$^{-1}$)</th>
</tr>
</thead>
<tbody>
<tr>
<td>1.33</td>
<td>0.0013</td>
<td>-4.70</td>
</tr>
<tr>
<td>1.15</td>
<td>0.0013</td>
<td>-5.16</td>
</tr>
<tr>
<td>0.85</td>
<td>0.0013</td>
<td>-5.70</td>
</tr>
<tr>
<td>1.33</td>
<td>0.0015</td>
<td>-5.26</td>
</tr>
<tr>
<td>1.15</td>
<td>0.0015</td>
<td>-5.75</td>
</tr>
<tr>
<td>0.85</td>
<td>0.0015</td>
<td>-6.24</td>
</tr>
<tr>
<td>1.33</td>
<td>0.0018</td>
<td>-6.00</td>
</tr>
<tr>
<td>1.15</td>
<td>0.0018</td>
<td>-6.52</td>
</tr>
<tr>
<td>0.85</td>
<td>0.0018</td>
<td>-7.00</td>
</tr>
</tbody>
</table>

Statistics returned by LINEST
-2630.7 2.029478 -4.03342
116.3984 0.120802 0.224514
0.992491 0.071753 #N/A
396.5193 6 #N/A
4.082931 0.030891 #N/A

The multiple regression equation,

$$\log \dot{\varepsilon} = 2.03 \log \sigma - 2631 \left( \frac{1}{T} \right) - 4.033$$  \hspace{1cm} (42)

in power-law form for the creep deformation is,

$$\dot{\varepsilon} = 10^{-4} \sigma^{2.03} \exp \left( \frac{-50}{RT} \right)$$  \hspace{1cm} (43)

i.e., $n = 2.0 \pm 0.1$, $A = 10^{-4}$ and $Q = 50 \pm 3$ kJ/mol. The regression values of $n$ and $Q$ are close to the experimental values. For the regression statistics, $r^2 = 0.9925$, F-statistic = 397, degree of freedom = 6; and the respective standard errors on $n$, $Q$ and $A$ are $\pm 0.1$, $\pm 3$ kJ/mol, and $\pm 0.2$, while the regression sum of squares = 4.1, residual sum of squares = 0.03, and standard error for the log $\dot{\varepsilon}$ estimate is $\pm 0.07$. The high value of F-statistics implies that the observed relationship between the dependent and independent variables did not occur by chance.

7. CONSTITUTIVE EQUATIONS FOR BOUNDARY CREEP MODEL

7.1. Superposition of Rate equations

For creep in a regime specified by $\sigma$ and $T$, the strongest obstacles control the deformation process. When several strengthening mechanisms (drag, discrete obstacles, lattice resistance, etc.) operate simultaneously, the contributions to the deformation are superimposed in approximately linear way.
7.1.1. Coupling Diffusional Flow and Power-law Creep

Plastic flow of fully-dense solid is caused by shearing, and the shear stress $\tau$ induces movement of defects (dislocations and vacancies) producing shear strain $\gamma$ and strain rate $\dot{\gamma}$ on the glide plane. For simple tension creep of polycrystalline Cu, $\tau$ and $\dot{\gamma}$ are related to the applied tensile stress $\sigma$ and strain rate $\dot{\varepsilon}$ by the relation [13],

\[
\begin{align*}
\tau &= \sigma / \sqrt{3} \\
\dot{\gamma} &= \frac{\dot{\varepsilon}}{3} \\
\end{align*}
\]  
(44)

The stress required to induce creep deformation is dependent on the strength of the microstructural defects and occurs via thermally activated process. If shear strain rate is the independent variable, then each mechanism of deformation is described by a rate equation relating the shear stress on the glide plane, temperature and structure of the material as,

\[
\dot{\gamma} = f(\tau_s, T, S, P)
\]

where $S$ describes the current microstructural state of the material, and $P_i$ are material properties (lattice parameter, atomic volume, bond energies, moduli, diffusion constants, etc.), assumed as constant. The state or microscopic variables $S_i$ normally change as the deformation progresses, and a second set of equations required for the rate of change is

\[
\frac{dS_i}{dt} = g_i(\tau_s, T, S, P)
\]

where $t$ is time. The coupled set of eqns. (45) and (46) are the constitutive law for the deformation mechanism, which upon integration over time, give the strain after any loading history. In steady state conditions, $\frac{dS_i}{dt} = 0$.

As the power-law creep rate ($\dot{\gamma}_p$) and diffusional flow rate ($\dot{\gamma}_d$) are independent deformation mechanisms involving different defects, to a first approximation, the strain rates are added, and the net creep rate of the polycrystal under stress $\tau_s$ at temperature $T$ becomes,

\[
\dot{\gamma}_{\text{net}} = \dot{\gamma}_p + \dot{\gamma}_d
\]

(47)

Within a regime of $\tau_s$ and $T$, the contribution of dominant mechanism to $\dot{\gamma}_{\text{net}}$ will be largest.

For tensile creep of Cu, where $n = 2$ and $Q < Q_k$, grain boundary sliding and diffusional flow each describes the sum of two additive contributions. Transition from viscous flow to power-law creep is used to find if the creep regimes are caused by independent deformation mechanisms or of one mechanism only. The assumptions of two independent mechanisms (Cobble and Power Law) provide better description of the observed creep behaviour of Cu, since $n = 2$ and $Q_k < Q_l$.

7.1.2. Macroscopic and Microscopic Relationship

To link experimental parameters with constitutive model parameters, the assumption is that plastic strain rate on slip plane follows Arrhenius type equation, which is applied in two different ways to analyse the creep data. The stress is presumed to influence either only the pre-exponential factor or only the activation energy term. In the first case, the rate eqn. (1) corresponds to eqn. (10), and with no implicit stress or temperature dependence leading to congruence between model parameters ($n$, $Q_k$) and experimental parameters ($n$, $Q_l$). The stress dependence on creep rate is determined by the activation volume, as indicated by the curve of hyperbolic shapes in Fig. 8, which suggests an inverse relation of $\Delta V_p \propto \sigma^{-1}$.

7.2. Boundary Creep Model Equations

7.2.1. Power-law Creep

At high temperatures (~ 0.5T_m), polycrystalline solid creep, and after the primary creep (transient) during which the state variables change, a steady state is reached in which the solid continued to deform with no significant change in the state variables.

Using the defining eqn. (44) and power law eqn. (10), the experimental parameters $Q_k$ and $n$, are related to the model or true parameters of activation free energy $\Delta G$, activation enthalpy $\Delta H$ and activation area $\Delta a$ through Arrhenius type shear strain rate equation [13],

\[
\dot{\gamma} = C \sigma^n \exp \left( - \frac{\Delta H(T)}{kT} \right) \\
\]  
(44)

The model activation energy $Q_k$ is expressed by

\[
Q_k = -R \frac{\partial \ln \dot{\gamma}}{\partial (1/T)} + \Delta H
\]

(49)

and the shear stress exponent $m$ by

\[
m = \left( \frac{\partial \ln \dot{\gamma}}{\partial (1/T)} \right) + \frac{\Delta W}{kT}
\]

(50)

where $\Delta W = \tau b \Delta a$ is the work done on the slip plane. If in eqn. (7), $\Delta H = \Delta Q$, then $m$ corresponds to $n$.

The rate equation for power-law creep can also be represented in the form [13],

\[
\dot{\gamma}_p = D_{\text{core}} \sigma^n \left( \frac{T}{\mu} \right)^{3/2}
\]

(51)

where $D_{\text{core}}$ is determined by the core diffusion
and grain boundary diffusion coefficients. At high T and low τ, lattice diffusion is dominant (HT creep), and the stress exponent is n; but at lower temperatures or higher stress, core diffusion becomes dominant (LT Creep) and the strain rate varies as τ^n+2 (exponent becomes n+2). The creep deformation of face-centred cubic metals, such as Cu, is influenced by stacking fault energy [28], leading to creep stress exponent > 3.5 for power-law; which suggests that some important physical quantities might be missing from the constitutive model equation.

At high stresses (> 10^3 μ), power-law breakdown occurs, as the measured strain rates are greater than predicted by eqn. (51), because the process is a transition from climb controlled to glide-controlled flow. Also the activation energy measured in the power-law breakdown regime gives values which exceed lattice self-diffusion energy, indicating that the recovery process differed from that of climb controlled creep. The observation reflects the temperature dependence of the shear modulus, which has greater effect when the stress dependence is greater for the exponential term.

7.2.2. Coble or Boundary Diffusional Flow

During creep, the applied stress changes the chemical potential ϕ of atoms at the surfaces of grains in a polycrystal, introducing potential gradient Δϕ, which at high temperatures induces diffusive flux of matter through and around the surfaces of the grains. The flux leads to strain when coupled with sliding displacement in the plane of the boundaries, which is diffusion controlled. At intermediate temperatures, the diffusional flow rate equation for Coble creep by grain boundary diffusion is [13],

\[ \dot{\gamma}_d = \frac{42Ωt_{eq}}{kTd^3} D_e \]  

(52)

where D_e is related to the grain size, boundary diffusion coefficient and effective thickness of the boundary. However, eqn (52) neglects the kinetics involved in detaching vacancies from grain boundary sites and reattaching again.

7.2.3. Grain Boundary Sliding

For creep tests at low stresses of σ/μ < 10^4 and intermediate temperatures of ~ 0.5 T_M, the deformation is attributed to Coble diffusion accommodated by grain boundary sliding. The grain boundary diffusion coefficient of Cu atoms is calculated from [29],

\[ D_{gbs} = \frac{d^3 kT}{150nw_Ω} \]  

(53)

where d ~ 100 μm, k = 1.38 x 10^{-23} J/K, ε = 2.2 x 10^{-5} s^{-1}, σ = 21.16 MPa, w is grain boundary width (~ 2b = 5.12 x 10^{-10} m) and Ω is the atomic volume (~ 1.18 x 10^{-29} m^3). The value of D_{gbs} = 6.3 x 10^{-14} m^2/s is higher than the lattice diffusion coefficient D = 1.7 x 10^{-18} m^2/s for self-diffusion of Cu atoms [26], and therefore Coble creep is dominant.

The movement of grain boundary is considered as a layer of width w sliding in a Newtonian viscous manner, so that the sliding rate on a planar boundary, \( \dot{Y}_{gbs} \) is given by [30],

\[ \dot{Y}_{gbs} = 16D_{gbs} \frac{μ}{kT} \left( \frac{b}{a} \right) \left( \frac{1}{n} \right) \]  

(54)

At small grain size, grain boundary sliding dominates at intermediate T = 0.5 T_M and low σ. Lifshitz grain boundary sliding is the necessary accommodating process for diffusional creep where n = 2, which is similar to creep of super-plastic materials.

7.3. Coupling Diffusional flow, Grain Boundary Sliding and Power law Creep

7.3.1. Accommodation of Coble creep

Creep deformation of polycrystalline Cu at intermediate temperatures and low stresses is attributed to Coble diffusion accommodated by grain boundary sliding and power law creep in the grain interior. If Creep creep rate (\( \dot{Y}_c \)), grain boundary sliding rate (\( \dot{Y}_{gbs} \)) and power law creep rate (\( \dot{Y}_p \)) are represented by the respective rate equations [13, 30],

\[ \dot{Y}_c = \frac{42Ωt_{eq}}{kT} D_e \left( \frac{1}{d^3} \right) \]  

(55)

\[ \dot{Y}_{gbs} = 16D_{gbs} \frac{μ}{kT} \left( \frac{b}{a} \right) \left( \frac{1}{n} \right) \]  

(54)

\[ \dot{Y}_p = AD_1 \frac{μ}{kT} \left( \frac{1}{n} \right)^2 \]  

(56)

then by superposition principle [31], the net creep rate is expressed in phenomenological form as,

\[ \dot{Y}_{net} = \dot{Y}_c + \dot{Y}_p + \dot{Y}_{gbs} \]  

\[ = \frac{42Ωt_{eq}}{kT} \left( \frac{1}{d^3} \right) + A \frac{μ}{kT} \left( \frac{b}{a} \right) \left( \frac{1}{n} \right)^2 + 16 \frac{D_{gbs} μ}{kT} \]  

(57)

In general form, eqn. (57) is equivalent to the rate equation of sum of independent thermally activated creep mechanisms,

\[ \dot{Y}_{net} = \sum_{i=1}^n \dot{Y}_i = \sum_{i=1}^n A_i \tau^{\theta_i} \exp \left( \frac{-\Phi_i}{RT} \right) \]  

(11)
7.2.4. Geometric Representation of Boundary flow

Due to the presence of diffusional flux of vacancies and consequent flow of matter to elongate the grains, relative movement of adjacent grains is required in order to maintain specimen coherency for the grain boundary sliding accommodated by plastic flow. As illustrated in Fig. 11(a), at the diffusion centres (marked with circles), the vacancy concentration gradient is zero. In grain (3), the build-up of material along BC leads to a velocity vector $V_1$ at the diffusion centre of the grain.

Fig. 11. Four grains in hexagonal array with relative movement (a), such that the total strain under tensile stress $\sigma$ is $\varepsilon = (\varepsilon_1 + \varepsilon_2)/L$ (b).

Similarly, there is a velocity vector $V_2$ due to the build-up of material at boundary AB. The diffusion centre of grain (3) therefore moves away from the triple point B, parallel to the tensile axis in the symmetric case, with a velocity $V$. Similar movement also occurs in the other grains, so that in the absence of any relative grain movement, the grains become elongated and assume shapes shown in Fig. 11(b).

Since coherency is assumed to be maintained during diffusion assisted boundary creep, the grains move relative to each other and a longitudinal marker XYZ is therefore divided into two segments $XY_1$ and $YZ$. The grain boundary sliding occurs as a result of grain movement accommodated by plastic flow through sequential processes of Newtonian flow and power law creep.

If the Cobble creep is dominant, the elongation of the grain structure results from vacancy flow along the grain boundary, since the activation energy is lower than energy for lattice self-diffusion. The flow of vacancies occurs from the grain boundary sliding perpendicular to boundaries parallel to the stress. In fine grained polycrystalline Cu, grain boundary sliding accommodates the dominant mechanism of Cobble creep, and results in large deformation since $n = 2$ [2, 32].

At elevated temperatures the grains in polycrystalline metals are able to move relative to each other. Boundary sliding is a shear process and occurs in the direction of the grain boundary or along the grain boundary as bulk movement of two grains, which is promoted by increasing the temperature or decreasing the strain.

8. DISCUSSION

Under the influence of applied stress of $\sigma/\mu < 10^4$, creep deformation occurs by flow of vacancies and interstitials through the crystal, while dislocation creep occurs in the stress range $10^4 < \sigma/\mu < 10^2$, and dislocation glide occurs under stress $\sigma/\mu > 10^2$. Diffusion creep is favoured at high $T$ and low $\sigma$, dislocation creep is more dominant at low $T$ and large $\sigma$, and diffusion creep becomes the controlling mechanism at high $T$ and relatively low $\sigma$.

At intermediate temperature $\sim 0.5T_M$, vacancy diffusion via grain boundary is favoured since activation energy for grain boundary sliding is typically $\sim 0.6Q_L$, and the creep rate is therefore determined by movement of the slower species along the fastest diffusion rate. If the lattice vacancy migration enthalpy of Cu is $\Delta H_1 = 80 \text{ kJ/mol}$, the migration enthalpy $\Delta H_g$ of $\sim 40 \text{ kJ/mol}$ for grain boundary is adequately provided by the activation energy of $60 \text{ kJ/mol}$ required to overcome the energy barrier for creep deformation.

When polycrystalline Cu is subjected to a tensile stress, defects migrate to the grains along the direction of tension, causing net mass transfer that elongates the crystal in the direction of maximum tension. Lifshitz grain boundary sliding occurring by diffusion in a low stress state relates to the boundary offset due to stress directed diffusion of atoms and vacancies in grain boundary as shown in Fig. 11. During sli-
ding, the grain boundary is assumed to be an extremely viscous thin layer between two grains, and the movement of the grain boundary is related to the diffusion processes [7, 8].

Reported values of n and Q for creep of Cu indicate that rate controlling mechanism by the intersection of obstacles (i.e. obstacle-controlled dislocation glide) [9, 10] is not applicable to creep deformation where comparatively low stresses are used and the average grain size of the specimen is also much smaller.

Grain boundary motion is often strongly influenced by small amounts of impurities even in high purity metals [11], and such situation could present a further difficulty in the study of the creep process. If the grain boundary motion is achieved by atoms moving from ledges at one crystal surface to the ledges on the other crystal surface, then vacancy diffusion parallel to the grain boundary would be required for accommodation.

9. CONCLUSIONS

Creep deformation of polycrystalline Cu at intermediate temperatures and low stresses is attributed to Coble diffusion accommodated by grain boundary sliding. In the absence of lattice resistance, the activation energy is sufficient to overcome the energy barrier for the deformation, and the steady state creep rate is a function of both applied stress and temperature.

The mean apparent stress exponent \( n_a = 2 \) at normalized stress \( < 10^4 \) and activation energy of 60 kJ/mol over temperature range \( T/T_M = 0.4 - 0.5 \) indicate that the dominant mechanism for the creep deformation is grain boundary movement. The value of the true stress exponent equals the apparent stress exponent, indicating that internal resistance to the creep deformation is negligible and the threshold stress has zero value. The apparent activation energy for the creep is lower than both lattice self-diffusion energy and true activation energy.

For intermediate temperature and low stress creep, the data is interpreted in terms of Coble diffusional creep for small grain size where \( n = 2 \) and \( Q = 0.6 \) \( Q_t \), accommodated by grain boundary sliding and power law creep. The grain boundary sliding occurs as a result of grain movement at the boundary and plastic flow in the grain, through sequential processes of Newtonian flow and power law creep.

A LINEST function multiple regression analysis correctly predicts the power-law creep equation for polycrystalline Cu, and the regression parameters agree with the experimental data.

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