



Review Article

Low temperature creep plasticity[☆]



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ABSTRACT

The creep behavior of crystalline materials at low temperatures ($T < 0.3T_m$) is discussed. In particular, the phenomenological relationships that describe primary creep are reviewed and analyzed. A discussion of the activation energy for creep at $T < 0.3T_m$ is discussed in terms of the context of higher temperature activation energy. The basic mechanism(s) of low temperature creep plasticity are discussed, as well.

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Michael E. Kassner graduated with a Bachelor in Science-Engineering from Northwestern University in 1972, and an M.S. and Ph.D. in Materials Science and Engineering from Stanford University in 1979 and 1981. Kassner accepted a position at Lawrence Livermore National Laboratory in 1981 and was employed there until 1990 where he was Head of the Physical Metallurgy and Welding Section. He accepted a faculty position

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the University of Southern California (USC). He is also a Professor of Materials Science at USC. He is currently Choong Hoon Cho Chair and Professor. He is currently active in pursuing research at USC on creep, fracture, fatigue and thermodynamics. Most recently, he was assigned to Washington, D.C., as the Director of Research at the Office of Naval Research (ONR). He assumed the position in October 2009 until October 2012. He was responsible for overseeing the nearly one billion dollar basic-research budget for the US Navy. He was awarded the Navy's Meritorious Public Service Medal for his tenure at ONR. He has published three books, one on the fundamentals of creep plasticity in metals, hot deformation of aluminum and aluminum alloys and another on phase diagrams and has authored or co-authored over 220 published articles. He is a Fellow of American Society of Metals (ASM), a Fellow of the American Society of Mechanical Engineers (ASME) and a Fellow of the American Association for the Advancement of Science (AAAS).

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Kamia Smith graduated from Wheaton College in Norton, Massachusetts in 2011 with a Bachelor of Arts degree in Chemistry. Smith then pursued graduate level studies at the University of Southern California. She completed her Masters of Science in Materials Materials Science and Engineering in 2013 and is now currently pursuing her PhD in the same field. Her research, under Dr. Michael Kassner, studies the mechanical behaviors of materials under various stress conditions.

During her first year as a doctorate student, Smith accepted a fellowship position in a program called Body Engineering Los Angeles GK-12. As a fellow, she worked in a Downtown Los Angeles middle school in which she taught her research to students in order to spark more engineering interest in younger generations. Smith looks to complete her studies by 2016.

1. Introduction

Temperature ranges for creep can be subdivided into three categories: (1) high temperature creep ($T > 0.6T_m$), (2) intermediate temperature creep ($0.3T_m < T < 0.6T_m$), and (3) low temperature creep ($T < 0.3T_m$). Generally, creep studies investigate high temperature deformation; however, this paper reviews on the latter category. Less attention has been paid to low temperature creep due to the fact that materials generally neither fail nor experience significant plasticity at lower (especially ambient and cryogenic) temperatures.

Creep at low temperature can be understood as time-dependent plasticity that occurs at $T < 0.3T_m$ and at stresses often below the macroscopic yield stress ($\sigma_y^{0.002}$). This is where creep is often not expected. Still, even with the low attention paid to this area of creep, many materials do experience very noticeable plasticity at lower temperatures. This has some commercial importance. These materials include Ti alloys and steels [1–10], Al–Mg [11], α -Brass [12], ionic solids [13], pure Au, Cd, Cu, Al, Ti, Hg, Ta, Pb, Zn [14–28] and precipitation hardened alloys [29], and glass and rubber [28].

Low-temperature creep has generally been investigated because of two reasons: (1) Materials may undergo plasticity that affects its intended performance. This category includes structural alloys, and creep of Cu at cryogenic temperatures; (2) There has been theoretical curiosity regarding low temperature deformation and the mechanism of plasticity, particularly at cryogenic temperatures. This includes the validity of the dislocation intersection mechanism proposed by Seeger et al. [30,31] as investigated by others [19]. Also there have been investigations of the proposition of quantum mechanical tunneling of dislocations at very low temperature [13,20,22,24,26,27].

1.1. Phenomenology

Generally, but not always, low temperature creep is a discussion of primary creep without the observation of a genuine mechanical steady state. One study has suggested steady state at 4.2 K, but there were problems with the data analysis [32].

At high temperatures, primary creep is described by the equations:

$$\varepsilon = \beta t^{1/3} + c_1 \quad (1)$$

as suggested long ago by Andrade [33] and Orowan [34]. Evans and Wilshire [35] reviewed the high-temperature primary creep equations and suggested a refinement. This refinement led to an equation of the form:

$$\varepsilon = at^{1/3} + ct + dt^{4/3} \quad (2)$$

This is now the common phenomenological equation used to describe primary creep. Variations to this equation include [36]:

$$\varepsilon = at^{1/3} + ct \quad (3)$$

and [37],

$$\varepsilon = at^{1/3}bt^{2/3} + ct \quad (4)$$

or,

$$\varepsilon = at^b + c^t \quad (5)$$

where [38],

$$0 < b < 1$$

or [1],

$$\varepsilon = at^b \quad (6)$$

where,

$$0 < b < 1$$

It is suggested that Eqs. (1)–(6) are all of a similar (power law) form. Another form of equations was suggested by Phillips [28], Laurent and Eudier [39] and Chévenard [40],

$$\varepsilon_p = \alpha \ln t + c_2 \quad (7)$$

Wyatt [18], long ago, suggested for pure metals, such as Al, Cd and Cu, that at higher temperatures, Eq. (1) was the proper descriptive equation, but at lower temperatures, he then suggested Eq. (7) was the proper form.

1.2. Objectives

The following discussion will describe the phenomenological trends in greater detail. The data appears to best be presented/described by material category (e.g. alloy, metal or ceramic). In particular, the low-temperature creep behavior of both alloys and pure metals will be described in separate sections. It will be shown that, generally, the descriptive equations generally fall within the forms of Eq. (1) or Eq. (2). Distinctions will be made for cases where the applied stress is above and below the conventional yield stress (at an ordinary strain-rate; e.g. 10^{-4} s^{-1}), as well as at shorter times than a few hours and much longer times.

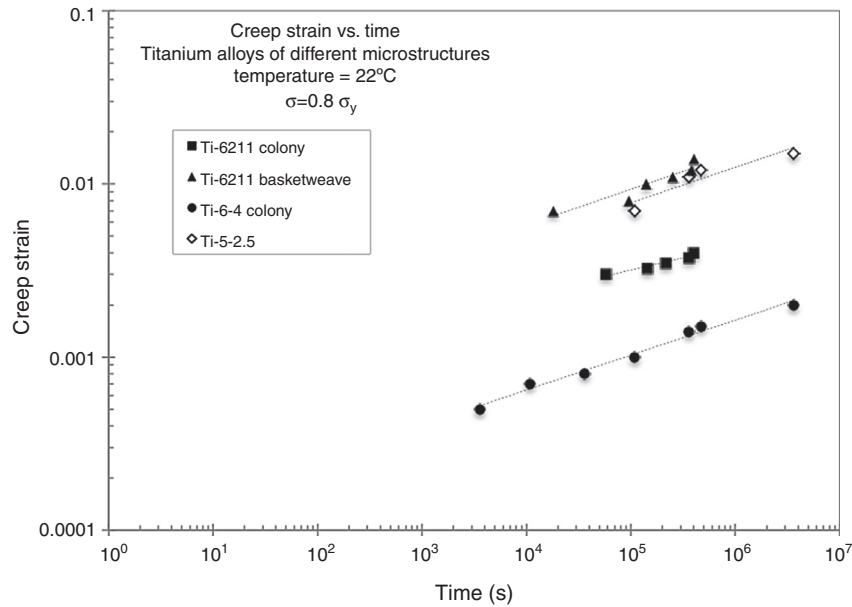


Fig. 1 – Ambient-temperature creep for various Ti-alloys of different microstructures and different compositions. From Neeraj et al. [1]. Power-law behavior is observed (Eq. (6)).

2. Temperature creep behavior of various metals and alloys

2.1. Titanium alloys

A useful discussion of low temperature creep in Ti alloys was presented by Neeraj et al. [1]. Fig. 1 plots ambient temperature creep data [3,11,41] for Ti alloys with different microstructures and different compositions. A particular value of this data is the extension of creep to longer times (over a month). The data shows that Eq. (1), the power-law relationship, reasonably describes the data. In contradiction to some of the earliest phenomenological equations, the exponent of $1/3$ does not appear to be unique for the b -value. Fig. 1 suggests a b -value closer to 0.2 . Also, the applied stresses in this case are all below the macroscopic yield stress. As Neeraj et al. point out, other literature confirms that $0.03 < b < 1$. A relatively recent effort to theoretically justify a value for $b = 1/3$ by Cottrell [42–44] and Nabarro [45] was performed.

2.2. Steels

2.2.1. AISI 4340 steel

Oehlert and Atrens [5] performed creep studies at ambient temperature, where for all creep conditions, the applied stress is below the yield stress. One of the tests was conducted at an applied stress of just half the yield stress. The durations of the tests were relatively short, and the times are only up to 20 min. These creep curves were described by a modification to Eq. (7):

$$\varepsilon_p = \varepsilon_{py} + \beta(\ln t) \quad (8)$$

where ε_{py} is the plastic strain on loading that is a function of stress. One clumsiness with Eq. (8) is that, at $T=0$, infinite creep rate is predicted. This was eliminated by modifying the equation to:

$$\varepsilon_p = \varepsilon_{py} + \alpha \ln(1 + \phi t) \quad (9)$$

where α and ϕ are constants. Oehlert and Atrens found that over the range of stresses, the constant α varied by a factor of nearly 10 and β by a factor of 2. The authors also examined 3.5 NiCr MoV and AeroMet100 with similar results.

2.2.2. 304 stainless steel

Fig. 2 shows the creep curves for annealed 304 stainless steel [7], that evinces some typical features of low temperature creep. This was discussed in the previous section. Fig. 2(a) shows the creep plot of the total plastic strain, ε_p , vs. time, t , for annealed 304. Fig. 2(b) shows the semi-log plot that illustrates the full time range of data [7]. Tests are conducted both above and below the yield stress. Tests above the yield stress were generally for a shorter term and strain was measured using an extensometer. However, below the yield stress, strain was measured for longer times using an optical comparator. It appears that the creep data for this material, at ambient temperature, follows a logarithmic behavior at a fixed stress, σ of

$$\varepsilon_p = \varepsilon_{py} + \beta(\ln t) \quad (10)$$

and this can be approximated by the usual relationship,

$$\varepsilon_{py} = a + b\sigma \quad (11)$$

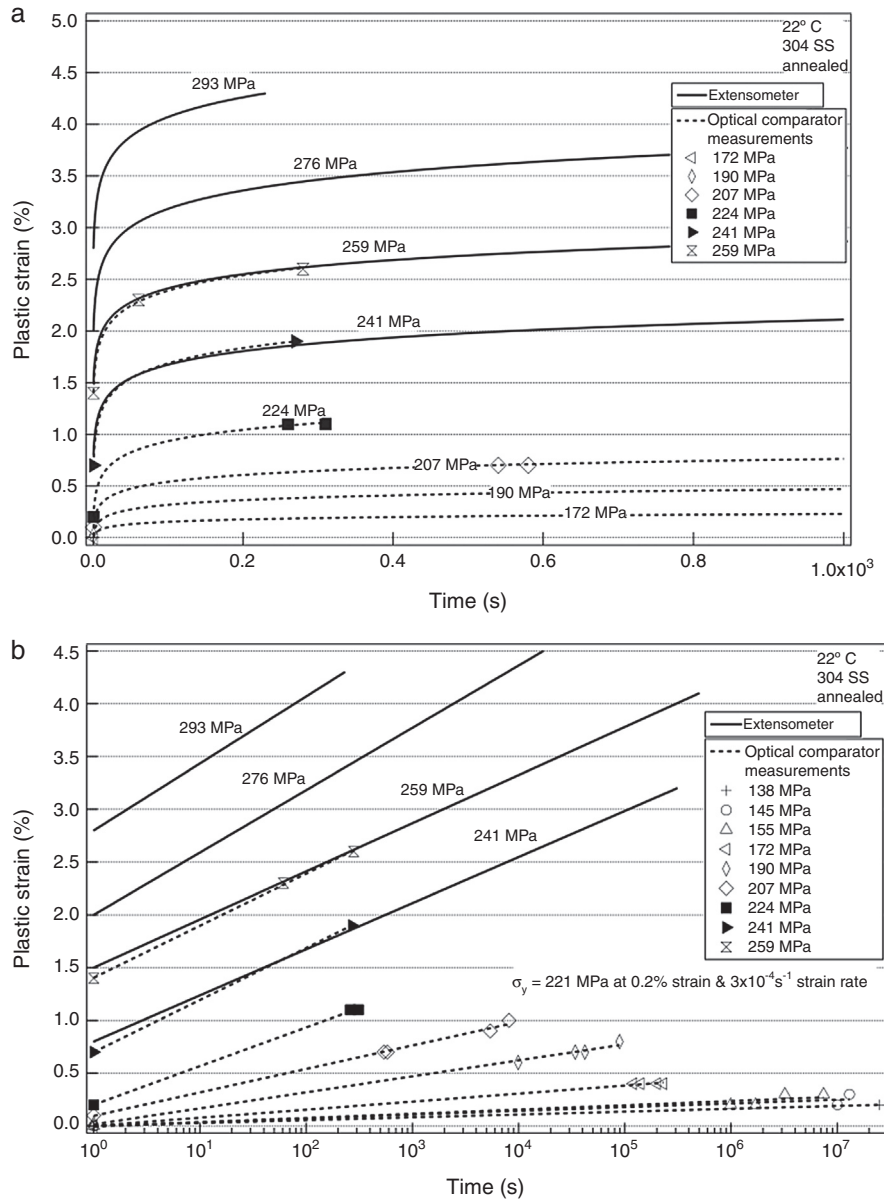


Fig. 2 – (a) The plastic strain vs. time behavior of annealed 304 stainless steel under different stresses. (b) Plastic strain vs. log time behavior of annealed 304 stainless steel [7]. Logarithmic behavior is observed (Eq. (7)).

where a and b are constants and a basically reflects the strain on loading. Also, β is the slope in Fig. 2(b) and appears to decrease with decreasing stress, approximated by,

$$\beta = -k\sigma + C_0 \quad (12)$$

where k and C_0 are constants.

Regression analysis suggests that for Eq. (11),

Above yield stress: $a = -8.12$ and $b = 0.037$
 Below yield stress: $a = -0.47$ and $b = 0.0027$

And for Eq. (12),

Above yield stress: $k = -0.0040$ and $C_0 = -0.52$
 Below yield stress: $k = -0.0028$ and $C_0 = -0.38$

2.3. Pure metals

2.3.1. Copper

Copper behavior is illustrated in Fig. 3 with data from Wyatt for shorter duration tests. Copper data for longer-term creep is also shown from Yen et al. [32]. The two sets of data are not in agreement and the explanation is unclear. The strains for the Yen et al. strain data appear low for the magnitude of stress greater than the yield stress, σ_y . Eq. (7) appears reasonable, except, possibly, for the lower two stresses of the Yen data which follow neither Eq. (1) nor Eq. (7).

2.3.2. Aluminum

The ambient-temperature creep behavior of high purity Al is illustrated in Fig. 4. Here, in contrast to Cu, power-law behavior

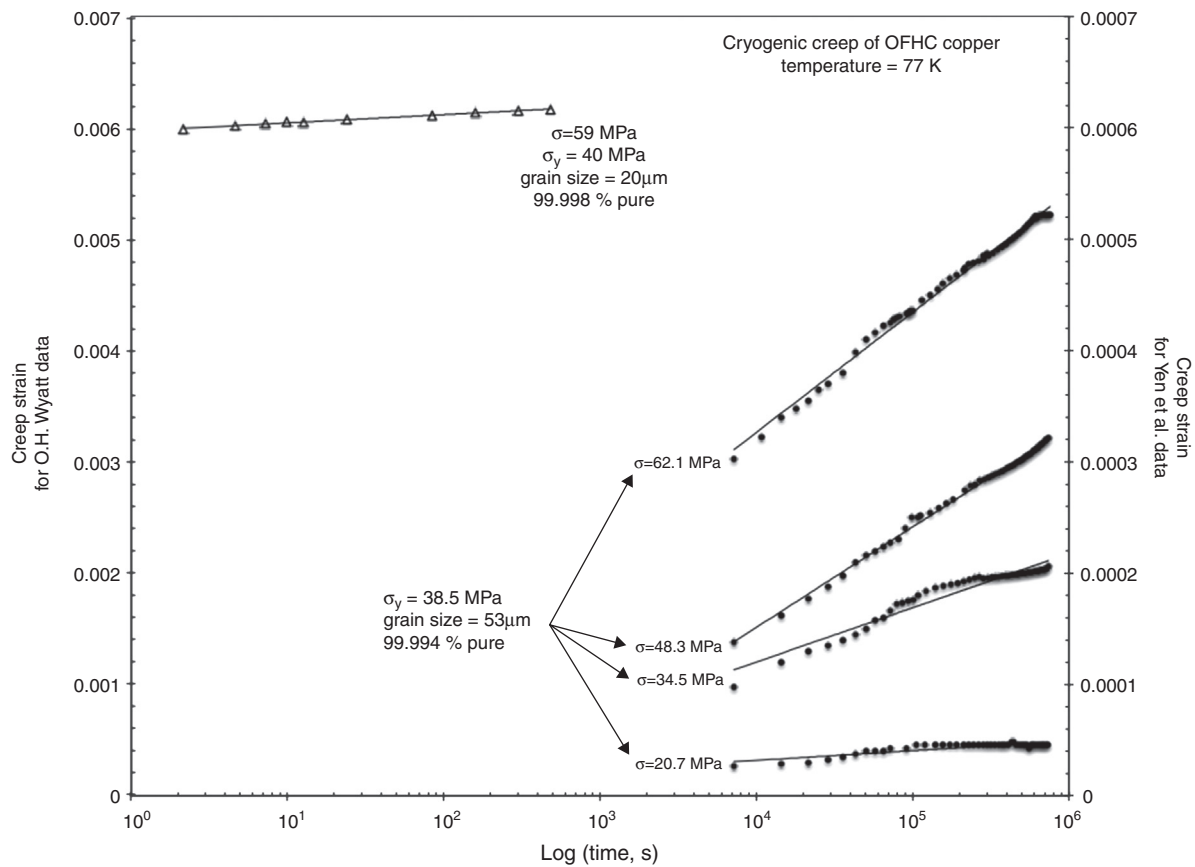


Fig. 3 – Creep strain vs. log time data for copper at various low temperatures and stresses [18,32]. Logarithmic behavior appears to dominate (Eq. (7)).

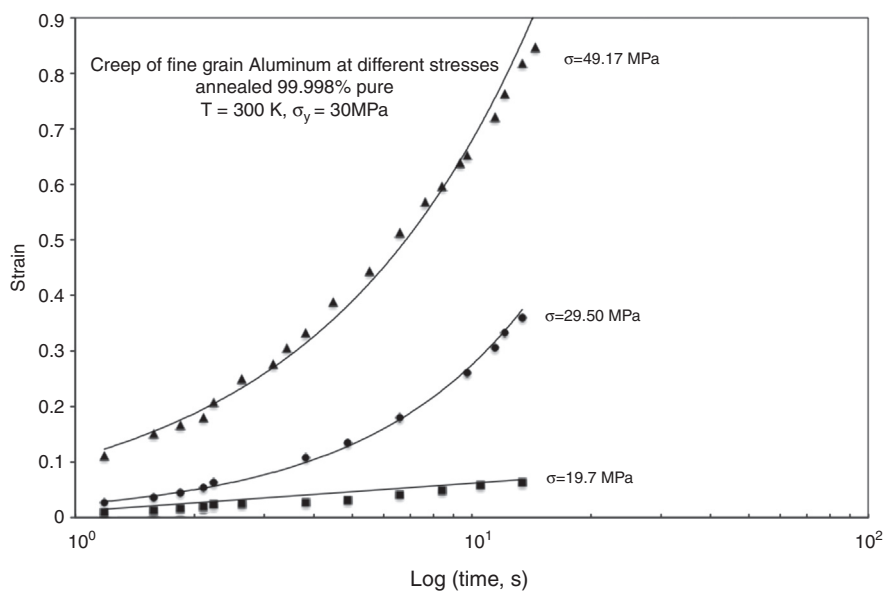


Fig. 4 – Ambient-temperature creep behavior pure Al based on data by [18]. Power-law creep is occurring but logarithmic behavior may be observed as well.

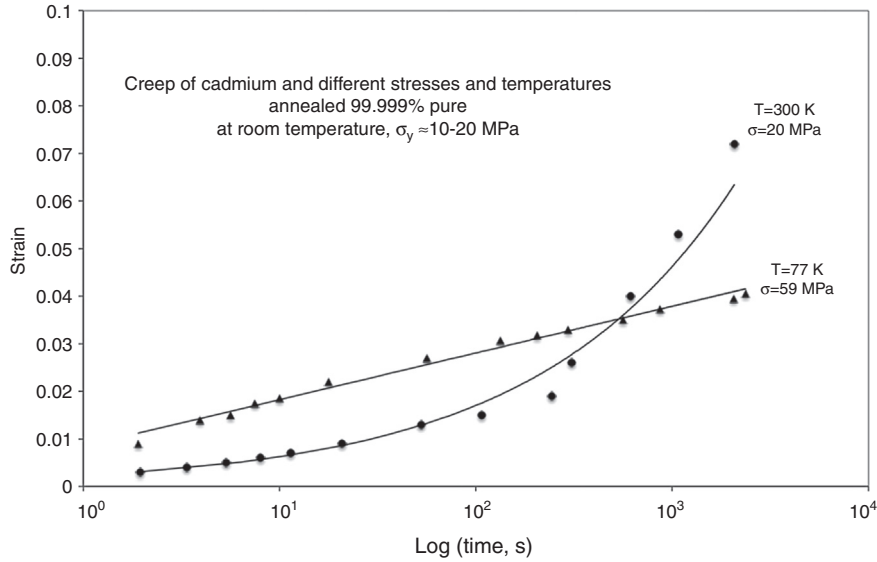


Fig. 5 – Ambient-temperature creep behavior of pure Cd based on data by [18]. Logarithmic creep may best describe the behavior, particularly at low temperatures and longer times.

(Eq. (6)) may dominate although Wyatt suggests logarithmic (Eq. (7)) may be influencing the behavior.

2.3.3. Cadmium

Fig. 5 illustrates the creep behavior of pure Cd at 77 K and at ambient temperature. It was concluded by Wyatt that Eq. (7) better describes the data at low temperature.

3. Mechanisms

3.1. Logarithmic creep

The earliest discussion of low temperature creep was based on dislocation glide and dislocation exhaustion by Mott and Nabarro [46]. There has also been substantial discussion of low temperature creep, especially low temperature plasticity, with published work on the dislocation intersection mechanism by Seeger et al. [30,31].

Here, the concept of activation volume and area were probably first used. The product of the activation area, A , Burger’s vector, b and the difference between the applied stress, τ and the back stress or long-range internal stress (LRIS), τ_G , due to other dislocations, is the energy supplied but is also the applied stress to allow a dislocation to surmount an obstacle. The activation area is usually defined as the product of the width of the obstacle, d , and the obstacle spacing, ℓ . Alternatively, the activation volume, v , equals $Ab = \ell db$. This leads to the classic rate equation [19,30],

$$\dot{\gamma} = NAbv_o \exp \left\{ \frac{-\Delta H_o - v(\tau - \tau_G)}{kT} \right\} \quad (13)$$

where $\dot{\gamma}$ is the strain-rate; N the number of dislocation segments per unit volume held up at the intersection points of mean spacing, ℓ . v_o is an atomic frequency of the order of the

Debye frequency; ΔH_o is the energy required for the intersection process, i.e. the energy for jog formation $\approx Gb^2/10$ [47].

$$\tau_G = \tau_G^o + \int_0^\gamma h d\gamma \quad (14)$$

Where, according to Conrad [19], τ_G^o is the stress due to the dislocations initially in the crystal and h the strain-hardening coefficient, which is defined as $d\tau/d\gamma$. It appears that $\tau_G \ll \tau$ for small strains [48].

Eqs. (13) and (14) suggest

$$\dot{\gamma} = NAbv_o \exp \left\{ \frac{-\Delta H_o - v(\tau - \tau_G^o - h\gamma)}{kT} \right\} \quad (15)$$

Integrating,

$$\exp \left\{ \frac{\Delta H_o - v(\tau - \tau_G^o - h\gamma)}{kT} \right\} d\gamma = (NAbv_o) dt \quad (16)$$

$$\left(\frac{kT}{vh} \right) \exp \left\{ \frac{\Delta H_o - v(\tau - \tau_G^o - h\gamma)}{kT} \right\} = NAbv_o t + D \quad (17)$$

$$\exp \left\{ \frac{\Delta H_o - v(\tau - \tau_G^o - h\gamma)}{kT} \right\} = (NAbv_o) \frac{vht}{kT} + D' \quad (18)$$

$$\frac{\Delta H_o - v(\tau - \tau_G^o - h\gamma)}{kT} = \ln(v't + D') \quad (19)$$

where,

$$v' = \left(\frac{NAbv_o v h}{kT} \right)$$

$$\left(\frac{hv\gamma}{kT} \right) = \ln(v't + D') - \frac{\Delta H_o + v(\tau - \tau_G^o)}{kT} \quad (20)$$

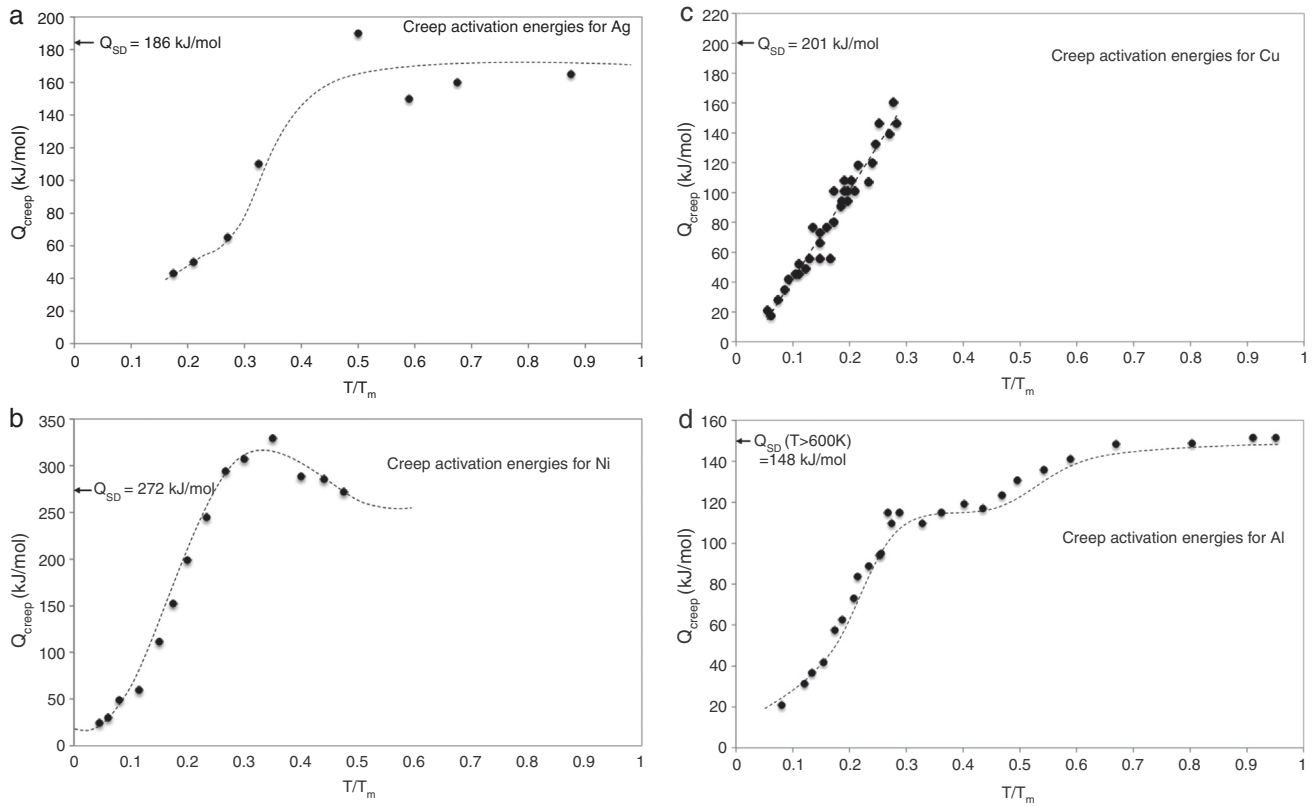


Fig. 6 – Activation energy for (steady-state) creep of Ag, Ni, Cu and Al as a function of temperature.

Source: Adapted from [49].

$$\gamma = \left(\frac{kT}{h\nu} \right) \ln(\nu't + D') - E \quad (21)$$

Choose a value for D' so that $\gamma(0) = 0$

$$\gamma = \left(\frac{kT}{h\nu} \right) \ln(\nu't + 1) \quad (22)$$

The last equation [19] is in the form of Eq. (7). Other attempts to justify this equation were made by Wyatt [18], but the methodology was unclear. Welch et al. [50] appeared to attempt a similar approach to that of Wyatt.

Other mechanisms considered include quantum mechanical tunneling, which predicts athermal creep behavior at low temperatures. Early proponents include [20,24,26,27]. Conversely, subsequent work by [13,22,23] suggests that, even to 4K, creep is time-dependent and is a result of the thermal activation of dislocations. Dislocation kink mechanisms were suggested for bcc metals [23]. However, it is unclear how this mechanism, by itself, explains the observed creep behavior. In materials with solutes, it is suggested that thermal activation past pinning solutes [11] is rate controlling such as with Al-Mg.

Increasing creep resistance at low temperatures appears to be accomplished in similar ways as at elevated temperatures. Where cold work increased the creep resistance in 304 stainless steel at elevated temperatures, it also increases creep resistance at low temperatures [51]. Others have suggested the role of other features [3] such as twin boundaries. Of course,

solute strengthening is a variable, such as in the 304 stainless steel. The precise mechanism by which the strengthening variables superimpose is unclear.

3.2. Power-law analysis

In an empirical analysis, Neeraj et al. [1] assume the Holloman flow equation:

$$\sigma = K\varepsilon^n \dot{\varepsilon}^m \quad (23)$$

where K is the strength parameter, n is the strain-hardening exponent, and m is the strain-rate sensitivity exponent. This equation was suggested to reasonably describe some Ti-alloy behavior. Thus, they show:

$$\dot{\varepsilon} = \frac{d\varepsilon}{dt} = \left(\frac{\sigma}{K} \right)^{1/m} \varepsilon^{-n/m} \quad (24)$$

$$\int \varepsilon^{n/m} = \int \left(\frac{\sigma}{K} \right)^{1/m} dt \quad (25)$$

under constant stress,

$$\varepsilon = \left(\frac{\sigma}{K} \right)^{1/m+n} \left(\frac{m+n}{m} \right)^{m/m+n} t^{m/m+n}$$

This is similar to Eq. (6),

$$\varepsilon = at^b$$

with

$$b = \left(\frac{m}{m+n} \right) \quad (26)$$

$$a = \left(\frac{\sigma}{K} \right)^{1/m+n} \left(\frac{m+n}{m} \right)^{m/m+n} \quad (27)$$

These equations relate constant strain-rate behavior to constant stress behavior. These investigators show interesting predictions of the low temperature creep behavior in Ti-alloys from a limited set of constant strain-rate data. They rationalized low temperature creep of Ti-alloys as being due to low n and moderate m values. This rationale may be extended to steels. Certainly, as also pointed out by [52], low n and high m certainly predisposes a material to significant creep plasticity at low temperatures.

3.3. Activation energies

Fig. 6 [49,53] shows the work that the activation energy decreases with temperature below $0.3T_m$ and is much lower than at higher temperatures where Q is associated with lattice self-diffusion.

Conflicts of interest

The authors declare no conflicts of interest.

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