Review Article

Creep in amorphous metals

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ABSTRACT

This paper reviews the work on creep behavior of amorphous metals. There have been, over the past several years, a few reviews of the mechanical behavior of amorphous metals. Of these, the review of the creep properties of amorphous metals by Schuh et al. though oldest of the three, is particularly noteworthy and the reader is referred to this article published in 2007. The current review of creep of amorphous metals particularly focuses on those works since that review and places the work prior to 2007 in a different context where new developments warrant.

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Michael E. Kassner graduated with a Bachelor in Science-Engineering from Northwestern University in 1972, and an MS and PhD 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 in the Mechanical Engineering Department at Oregon State University in 1990 where he was Northwest Aluminum Professor of Mechanical Engineering. He received the College of Engineering Outstanding Sustained Research Award in 1995. He moved in 2003 to accept a position as Chairman, Mechanical and Aerospace Engineering Department at 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, DC, 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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Amorphous properties of bulk metal 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

Amorphous metals are a relatively new class of alloy, originating in about 1960 with the discovery of thin metallic ribbons by splat cooling [1]. These are always alloys, and pure metal glasses have not yet been produced. As these alloys are non-crystalline, they have no dislocations, at least in the sense normally described in crystalline materials. Thus, amorphous metals have yield stresses that are higher than crystalline alloys. High fracture stress, low elastic modulus, and sometimes-favorable fracture toughnesses are observed. Often, favorable corrosion properties were observed, as well, partly due to an absence of grain boundaries. Towards 1990, alloys with deep eutectics were developed that allowed liquid structures to be retained in thicker sections in the amorphous state on cooling to ambient temperature [2–13]. With this development, there has been fairly intensive study of bulk metallic glasses (BMG) for possible structural applications. Most of the alloys in this chapter are relevant to BMGs. Table 1 lists some of the short-term mechanical properties of some BMGs taken from [14–34] and some of the impressive properties are listed.

Fig. 1, based on an illustration by [35], is a time–temperature-transformation (T–T–T) diagram that illustrates some of the important temperatures for metallic glasses. First, there is the equilibrium liquid to solid transition at the melting temperature Tm where, of course, multiple solid crystalline phases form on cooling. Below this temperature a T–T–T curve is illustrated. Cooling below Tm must be sufficiently rapid to avoid intersecting the “nose” of the curve. Also illustrated is the glass transition temperature, Tg.

This is generally assigned to that temperature where there is discontinuity in the change of a property (e.g. heat capacity, thermal expansion coefficient, etc.) with temperature. The region between Tm and Tg is generally referred to as the supercooled liquid regime. Some values for various BMGs are listed in Table 2 [36–47].

The discussions in this chapter will be largely confined to temperatures above 0.7 Tg. As will be discussed subsequently, this is the regime in which homogeneous deformation is observed. This review refers to this regime as a “creep regime” of amorphous metals. A practical importance of this regime is that this is where forming of a metallic glass is frequently performed. This regime is contrasted by the regime of lower temperatures where heterogeneous deformation or shear banding is often (but not always) observed.

2. Mechanisms of deformation

2.1. Overview

The suggested deformation mechanism has generally fallen into three categories: (a) Dislocation-like defects [48–50], (b) diffusion-type deformation [51], and (c) shear transformation zones (STZs) [52,53]. These are illustrated in Fig. 2, and are all early explanations for plasticity, but it appears that the amorphous metals community has generally embraced the third, STZ [12,13,54].

The essence of this latter mechanism is that there is a so-called “free volume” in amorphous metals. Free volume is a “concept” and it has no absolute definition. The starting state is the baseline; only the difference has meaning, so a change in density after deformation defines the free volume. The exact form and shape of these free volumes is not known. Increasing free volume would be associated with decreased density. Estimates for free volume for Zr41.2Ti13.8Cu12.5Ni10Be22.5 (Vitreloy 1) is about 3% [54]. Free volume decreases (tighter packing) appear to increase ductility in homogeneous deformation

**Fig. 1** – A time–temperature-transformation diagram that illustrates the important temperature regions of BMGs from Ref. [36].
Table 1 – Mechanical properties of some glassy alloys from Ref. [13].

<table>
<thead>
<tr>
<th>Material</th>
<th>$E$ (GPa)</th>
<th>$\sigma_f$ (MPa)</th>
<th>$\sigma_y$ (MPa)</th>
<th>$\epsilon_f$ (%)</th>
<th>$\epsilon_y$ (%)</th>
<th>References</th>
</tr>
</thead>
<tbody>
<tr>
<td>Co$<em>{43}$Fe$</em>{26}$Ta$<em>5$Sb$</em>{15}$</td>
<td>268</td>
<td>5185</td>
<td>2</td>
<td>[14]</td>
<td></td>
<td></td>
</tr>
<tr>
<td>Cu$<em>{50}$Hf$</em>{25}$Ti$_{15}$</td>
<td>124</td>
<td>2024</td>
<td>2088</td>
<td>1.6</td>
<td>[15]</td>
<td></td>
</tr>
<tr>
<td>(Cu$<em>{40}$Hf$</em>{25}$Ti$<em>{15}$)$</em>{10}$Nb$_{4}$</td>
<td>130</td>
<td>2405</td>
<td>2.8</td>
<td>[16]</td>
<td></td>
<td></td>
</tr>
<tr>
<td>Cu$<em>{42}$Ti$</em>{25}$Zr$<em>{11}$Nd$</em>{5}$Si$_{17}$</td>
<td>1930</td>
<td>2250</td>
<td>[17]</td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>Gd$<em>{40}$Zr$</em>{50}$</td>
<td>84</td>
<td>1272</td>
<td>1794</td>
<td>1.7</td>
<td>6.2</td>
<td>[18]</td>
</tr>
<tr>
<td>Co$<em>{46}$Zr$</em>{54}$</td>
<td>92.3</td>
<td>2000</td>
<td>2.2</td>
<td>[19]</td>
<td></td>
<td></td>
</tr>
<tr>
<td>(Fe$<em>{45}$Co$</em>{20}$)$<em>{46.4}$Mo$</em>{54}$Cu$<em>{15}$Er$</em>{0.5}$</td>
<td>192</td>
<td>3700</td>
<td>4100</td>
<td>0.55</td>
<td>[20]</td>
<td></td>
</tr>
<tr>
<td>Fe$<em>{77}$Nb$</em>{23}$</td>
<td>200</td>
<td>4850</td>
<td>6.6</td>
<td>[21]</td>
<td></td>
<td></td>
</tr>
<tr>
<td>Fe$<em>{25}$Si$</em>{75}$B$_{25}$</td>
<td>200</td>
<td>4200</td>
<td>2.1</td>
<td>1.9</td>
<td>[22]</td>
<td></td>
</tr>
<tr>
<td>Fe$<em>{87}$Mo$</em>{13}$C$_{5}$</td>
<td>3330</td>
<td>3400</td>
<td>2.2</td>
<td>[23]</td>
<td></td>
<td></td>
</tr>
<tr>
<td>Cu$<em>{42}$Fe$</em>{40}$Al$<em>{18}$Cu$</em>{5}$</td>
<td>210</td>
<td>4100</td>
<td>4250</td>
<td>2.2</td>
<td>[24]</td>
<td></td>
</tr>
<tr>
<td>Cu$<em>{42}$Fe$</em>{40}$Al$<em>{18}$Cu$</em>{5}$</td>
<td>220</td>
<td>3750</td>
<td>4140</td>
<td>0.25</td>
<td>[25]</td>
<td></td>
</tr>
<tr>
<td>Gd$<em>{40}$Co$</em>{50}$Al$_{50}$</td>
<td>70</td>
<td>1380</td>
<td>1.97</td>
<td>[26]</td>
<td></td>
<td></td>
</tr>
<tr>
<td>Ni$<em>{42}$Al$</em>{28}$B$_{18}$</td>
<td>3080</td>
<td>5</td>
<td>[27]</td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>Pd$<em>{77}$Cu$</em>{23}$</td>
<td>1476</td>
<td>1600</td>
<td>11.4</td>
<td>[28]</td>
<td></td>
<td></td>
</tr>
<tr>
<td>Pd$<em>{77}$Cu$</em>{23}$</td>
<td>82</td>
<td>1475</td>
<td>1575</td>
<td>[29]</td>
<td></td>
<td></td>
</tr>
<tr>
<td>Pd$<em>{25}$Cu$</em>{143}$Ni$_{25}$</td>
<td>1400</td>
<td>1470</td>
<td>2</td>
<td>[30]</td>
<td></td>
<td></td>
</tr>
<tr>
<td>Ti$<em>{15}$Zr$</em>{50}$Hf$<em>{35}$Cu$</em>{12.5}$Ni$<em>{17.5}$Si$</em>{17.5}$</td>
<td>95</td>
<td>2040</td>
<td>0</td>
<td>[31]</td>
<td></td>
<td></td>
</tr>
<tr>
<td>Zr$<em>{35}$Cu$</em>{65}$Al$<em>{50}$Ni$</em>{50}$</td>
<td>1410</td>
<td>1420</td>
<td>[32]</td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>Zr$<em>{41.5}$Ti$</em>{31.5}$Cu$<em>{12.5}$Ni$</em>{18}$Be$_{25.5}$</td>
<td>96</td>
<td>1900</td>
<td>1900</td>
<td>2</td>
<td>[33]</td>
<td></td>
</tr>
<tr>
<td>Zr$<em>{57}$Nb$</em>{25}$Al$<em>{12}$Cu$</em>{15}$Ni$_{6}$</td>
<td>86.7</td>
<td>1800</td>
<td>1800</td>
<td>2</td>
<td>[34]</td>
<td></td>
</tr>
</tbody>
</table>

Note: $E$, Young’s modulus; $\sigma_f$, yield strength; $\sigma_y$, fracture stress; $\epsilon_f$, elongation at yielding; $\epsilon_y$, plastic elongation. All the tests were conducted under compression, generally, at strain-rates from $(1-5) \times 10^{-4}$ s$^{-1}$.

Table 2 – Deformation data of some BMGs in the super-cooled liquid region from Ref. [78].

<table>
<thead>
<tr>
<th>Alloys (in at%)</th>
<th>$T_g$ (K)</th>
<th>$T_m$ (K)</th>
<th>$m$ value</th>
<th>Ductility*</th>
<th>References</th>
</tr>
</thead>
<tbody>
<tr>
<td>La$<em>{25}$Al$</em>{75}$Ni$_{10}$</td>
<td>480</td>
<td>520</td>
<td>1.0</td>
<td>1800</td>
<td>[36]</td>
</tr>
<tr>
<td>Zr$<em>{60}$Al$</em>{20}$Ni$<em>{20}$Cu$</em>{15}$</td>
<td>652</td>
<td>757</td>
<td>0.8–1.0</td>
<td>340</td>
<td>[37]</td>
</tr>
<tr>
<td>Zr$<em>{52}$Al$</em>{18}$Ti$<em>{18}$Cu$</em>{18}$Ni$_{15}$</td>
<td>458</td>
<td>456</td>
<td>0.45–0.55</td>
<td>650</td>
<td>[38]</td>
</tr>
<tr>
<td>Zr$<em>{55}$Cu$</em>{15}$Al$<em>{5}$Ni$</em>{15}$</td>
<td>683</td>
<td>763</td>
<td>0.5–1.0</td>
<td>N/A</td>
<td>[39]</td>
</tr>
<tr>
<td>La$<em>{50}$Al$</em>{15}$Ni$<em>{15}$Co$</em>{15}$</td>
<td>451</td>
<td>523</td>
<td>1.0</td>
<td>N/A</td>
<td>[40]</td>
</tr>
<tr>
<td>Pd$<em>{60}$Ni$</em>{20}$Pd$_{20}$</td>
<td>589</td>
<td>670</td>
<td>0.5–1.0</td>
<td>0.94</td>
<td>[41]</td>
</tr>
<tr>
<td>Zr$<em>{55}$Al$</em>{25}$Ni$<em>{25}$Cu$</em>{25}$</td>
<td>652</td>
<td>757</td>
<td>0.83</td>
<td>750</td>
<td>[42]</td>
</tr>
<tr>
<td>Zr$<em>{55}$Al$</em>{25}$Cu$<em>{25}$Ni$</em>{25}$</td>
<td>670</td>
<td>768</td>
<td>0.5–0.9</td>
<td>800</td>
<td>[43]</td>
</tr>
<tr>
<td>Ti$<em>{45}$Zr$</em>{45}$Ni$<em>{10}$Cu$</em>{5}$Be$_{15}$</td>
<td>601</td>
<td>648</td>
<td>N/A</td>
<td>1.0</td>
<td>[44]</td>
</tr>
<tr>
<td>Cu$<em>{50}$Zr$</em>{50}$Hf$<em>{50}$Ti$</em>{10}$</td>
<td>721</td>
<td>766</td>
<td>0.3–0.61</td>
<td>0.78</td>
<td>[45]</td>
</tr>
<tr>
<td>Zr$<em>{52.5}$Al$</em>{18}$Cu$<em>{18}$Ti$</em>{18}$Ni$_{15}$</td>
<td>659</td>
<td>761</td>
<td>0.5–1.0</td>
<td>&gt;1.0</td>
<td>[46]</td>
</tr>
<tr>
<td>Zr$<em>{41.5}$Ti$</em>{31.5}$Cu$<em>{12.5}$Ni$</em>{18}$Be$_{25.5}$</td>
<td>614</td>
<td>698</td>
<td>0.4–1.0</td>
<td>1624</td>
<td>[47]</td>
</tr>
</tbody>
</table>

* “T” and “C” stand for tension and compression, respectively.

at ambient temperature [55]. With an applied stress, groups of atoms (e.g. few to 100 [2,54,56]), under an applied shear stress, $\tau$, move and perform work. This constitutes an STZ. Argon et al. [51,52] considered that the STZ operation takes place within the elastic confinement of a surrounding glass matrix, and the shear distortion leads to stress and strain redistribution around the STZ region [2,51,52]. When the STZs exist throughout the alloy we have homogeneous deformation. STZs also occur in shear bands leading to heterogeneous deformation. STZs have been observed to create free volume during heterogeneous deformation [57,58]. Steady-state flow within the homogeneous regime can be considered a case where there is a balance between free volume creation and annihilation [77].

Fig. 2b illustrates the STZ mechanism. Argon et al. described the activation energy for this process and Schuh et al. estimated the predicted activation energy, Q, as 100–500 kJ/mol. Table 3 [52,56,59,60] lists some of the experimentally observed activation energies which are consistent with Argon’s predictions.

Table 3 – Activation energies for creep of selected metallic glasses ([56,59,60]).

<table>
<thead>
<tr>
<th>Composition</th>
<th>$T_g$ (K)</th>
<th>$T_m$ (K)</th>
<th>$\Delta Q$ (kJ/mol)</th>
</tr>
</thead>
<tbody>
<tr>
<td>Al$<em>{25}$Cu$</em>{25}$Zr$_{50}$</td>
<td>740</td>
<td>573</td>
<td>230.12</td>
</tr>
<tr>
<td>Cu$<em>{40}$Zr$</em>{60}$</td>
<td>677</td>
<td>543</td>
<td>218.82</td>
</tr>
<tr>
<td>Cu$<em>{50}$Zr$</em>{50}$</td>
<td>727</td>
<td>573</td>
<td>217.57</td>
</tr>
<tr>
<td>Cu$<em>{60}$Zr$</em>{40}$</td>
<td>750</td>
<td>573</td>
<td>228.45</td>
</tr>
<tr>
<td>Pd$<em>{60}$Si$</em>{40}$</td>
<td>673</td>
<td>546</td>
<td>191.63</td>
</tr>
<tr>
<td>Zr$<em>{35}$Cu$</em>{50}$Al$<em>{15}$Ni$</em>{15}$</td>
<td>410</td>
<td></td>
<td></td>
</tr>
<tr>
<td>Au$<em>{48}$Ag$</em>{5}$Pd$<em>{25}$Cu$</em>{25}$Si$_{16.3}$</td>
<td>103</td>
<td></td>
<td></td>
</tr>
<tr>
<td>Zr$<em>{41}$Ti$</em>{31}$Cu$<em>{12.5}$Ni$</em>{18}$Be$_{25}$</td>
<td>625/632</td>
<td>366</td>
<td></td>
</tr>
</tbody>
</table>
The equations that have been used to describe the creep-rate based on STZ have used the classic rate equation formalism leading to [2]:

\[ \dot{\gamma} = \alpha_0 \nu_0 \cdot \exp \left( -\frac{Q}{kT} \right) \sinh \left( \frac{\tau V}{kT} \right). \] (1)

where \(\alpha_0\) is a constant that includes the fraction of material deforming by activation, \(\nu_0\) is an attempt frequency, and \(\gamma_0\) is the characteristic strain of an STZ, and \(V\) is the activation volume. The hyperbolic sine function arises, as there can be both a forward and reverse “reaction”.

At low stresses \((\tau \ll kT/V)\), this equation reduces to the Newtonian:

\[ \dot{\gamma} = \frac{\alpha_0 \nu_0 \tau}{kT} \cdot \exp \left( -\frac{Q}{kT} \right) \tau. \] (2)

since “reverse” deformation is irrelevant.

Conversely, at stresses, \(\tau \gg kT/V\),

\[ \dot{\gamma} = \frac{1}{2} \alpha_0 \nu_0 \tau \cdot \exp \left( -\frac{Q}{kT} - \frac{\tau V}{kT} \right). \] (3)

Schuh et al. [2] point out that Eq. (1) suggests a Newtonian region followed by, with increasing stress, continual increase in stress exponent. Examples of BMGs that have evinced Eqs. (1)–(3) behaviors are illustrated in Figs. 3 and 4, which plot the steady-state creep behavior of several BMGs [2,36].

The figures on steady-state behavior illustrate that with increasing strain-rate and/or decreasing temperature there is a breakdown in Newtonian behavior and the apparent stress-exponent increases. Generally, this has been regarded as a natural consequence of Eq. (2), the rate equation that predicts Newtonian behavior at low stresses (higher temperature and lower strain-rates) but increased exponents with higher stresses (low temperatures and higher strain-rates). This explanation does not appear to be unanimously embraced [36]. For some, an important question is whether the non-Newtonian homogeneous deformation region is actually a reflection of nano-crystallization.

These equations suggest that free volume is largely responsible for plastic flow; larger free volumes would appear to more easily lead to regions of plastic flow. Schuh et al. point out that atomic simulations have suggested that other variables such as short-range chemical ordering can affect plasticity as well, which is not explicitly included in the above equations [61–64]. The pressure sensitivity of these equations was addressed by Sun et al. [65].

Nieh and Wadsworth [36] found that nano-crystallization occurred in Zr10Al5Ti19.9Cu14.6Ni BMG coincident with the deviation from Newtonian behavior. Nieh rationalized the nano-crystalline precipitates as akin to dispersion strengthening. Suryanarayana and Inoue [66] appear to suggest that the stress exponent increases due to second phase strengthening of the nanoparticles by a straightforward rule of mixtures for the flow strength. Schuh et al. [2] referenced the Nieh and Wadsworth work and certainly acknowledged the observation that deformation can induce crystallization (as have others [66–69]) but appear to favor the rate equation as an explanation for the deviation from Newtonian behavior at higher stresses. Wang et al. [70] found only non-linear creep behavior in Vitreloy 1 if some crystallization occurred. Whereas Newtonian conditions led to elongations in La55Al25Ni20 in excess of 20,000% [36], those at higher rates with non-Newtonian behavior exhibited dramatically reduced values. Many authors [2,12,13] proposed metallic glass deformation maps, similar to the construct by Ashby and Frost for crystalline materials. A metallic glass deformation–map is illustrated in Fig. 5.

Fig. 2 – (a) Two-dimensional representation of a dislocation line in crystalline (left) and amorphous (right) solids, atomistic deformation of amorphous metals in the form of (b). Shear transformation zones (STZ), and (c). Local atomic jump. Figure adapted from [12].
Newtonian deformation appears to reflect a fully amorphous alloy, but at least in other regions, including heterogeneous deformation, nano-crystallization may be occurring [66–69].

Furthermore, as will be discussed in a subsequent section, homogeneous deformation may extend to low temperature, in at least some cases.

2.2. Homogeneous flow at very low temperatures

Recent work [57,71–73] shows that, given sufficient time, homogeneous deformation can be detected under “electrostatic” (i.e. at a stress less than the yield stress, \(\sigma_y\)) loading at room temperature (RT). The stress exponent was not assessed, so it was unclear whether Newtonian flow was observed. Alloys include Zr_{46.75}Ti_{8.25}Cu_{17.5}Ni_{10}Be_{27.5}, Ni_{62}Nb_{38}, Cu_{50} Zr_{50}, Cu_{50} Zr_{43} and Cu_{65} Zr_{35}. Of course, some BMGs, such as Zn_{20}Cu_{20}Ta_{20}(Li_{40.55}Mg_{59.45})_{20}, may have low \(T_g\) (323 K) allowing homogeneous deformation at RT [74]. Alloys with higher packing densities exhibit greater plastic strain during homogeneous deformation at room temperature, but show less global plasticity during inhomogeneous deformation in a typical compression test [55]. Park et al. [57] suggest deformation induced structural disordering by molecular dynamic simulations, although others [73] imply STZ as the mechanism. Compression tests on Pd_{77}Si_{23} with cylindrical samples of diameters 8 μm and 140 nm showed that as the sample size decreased to the submicron range, homogeneous deformation occurs and was suggested to occur due to the necessity of a critical size volume for shear bands [75]. Similar results were noted by others [76].

2.3. Anelasticity

In the above discussion of the so-called “electrostatic” regime, a substantial fraction of the (small) non-elastic strain is anelastic. It should be noted that the STZ model naturally predicts some anelasticity. An isolated STZ, by the Argon model, is elastically constrained during activation. This implies that even at a low applied stress (where backflow, according to Eq. (2), is negligible), there is, nonetheless, a back stress that on unloading leads to anelastic back flow. It was additionally pointed out by Ke et al. [72] that there is a range of atomic environments in a glass such that some atoms reside in regions where the local topology is unstable. In these regions, the response to shear stress may include not only atomic displacements but also an anelastic reshuffling of the atomic near-neighbors (i.e. an anelastic STZ operation). Even though

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**Fig. 3** – Steady-state homogeneous flow data for Zr_{61.2}Ti_{13.8}Cu_{12.5}Ni_{10}Be_{22.5} metallic glass at elevated temperatures, from the work of Lu et al. [53]. Figure based on [2].

**Fig. 4** – Stress–strain rate curve for a Zr_{10}Al_{5}Ti_{17.2}Cu_{14.4}Ni glassy alloy shows Newtonian flow at low strain rates but non-Newtonian at high strain rates (data from Ref. [37]). Figure based on [36].
the fraction of atoms involved in these events may be small, the local strains are large enough that their cumulative effect makes a significant contribution to the macroscopic strain [72].

2.4. Primary and transient creep (non-steady-state flow)

Steady-state flow has principally been discussed. So far, it has been presumed that STZs create free volume (leading to softening) and that recovery processes involve the annihilation of free volume (leading to hardening). Therefore, steady state has been regarded as a balance between free volume creation and annihilations. Other hardening effects such as chemical ordering have not been explicitly considered for steady state. It has been suggested that there can be a net free volume increase or decrease during deformation that precedes a steady state. Fig. 6 from Lu et al. [53] shows hardening at the onset of deformation that continues beyond the eventual steady state. The interpretation of this peak stress followed by softening to a steady state is unclear.

3. Summary

This review of creep (above 0.7 $T_g$) in amorphous alloys emphasizes a variety of conclusions. First, the mechanism of creep appears to largely be explained by shear transformation zones where deformation is homogeneous. The descriptive equation for STZs suggests a Newtonian region followed by, with increasing stress, a continual increase in the stress exponent. It is not clear, as some suggest, that the non-Newtonian behavior is due to nano-crystallization. Second, homogeneous deformation at room temperature has recently been observed. Third, a substantial fraction of the small non-elastic strain at room temperature is anelastic. Fourth, during primary creep, STZs create free volume, leading to softening. Furthermore, recovery processes or annihilation of free volume leads to hardening.

**Conflicts of interest**

The authors declare no conflicts of interest.
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