Original Article

The effect of addition of Sn to copper on hot compressive deformation mechanisms, microstructural evolution and processing maps

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Abstract
The high-temperature compressive deformation behavior of a Cu-10 wt.% Sn solid solution alloy was studied in the temperature range of 843–993 K and in the strain rate range of 10^{-3} – 10^{-1}s^{-1} and the relationships among the flow stress, strain rate and temperature were determined. Based on the data obtained from the compression tests, the deformation mechanisms were identified, and the processing maps were constructed. The Cu-10Sn alloy exhibited solute drag creep at low strain rates and high temperatures and power law breakdown (PLB) at high strain rates and low temperatures. In the processing maps, above the temperature of 933 K, the alloy did not show flow instability up to a high strain rate of 10 s^{-1} and showed high power dissipation efficiencies (31–35%) even at the very high strain rate of 10 s^{-1}. Furthermore, at 993 K, the fraction of dynamically recrystallized grains after compressive deformation was as high as 0.83 at 10 s^{-1}. These results indicate that the Cu-10Sn alloy exhibits a significantly better workability and a higher quality of post-deformation microstructures compared to the pure Cu where dislocation climb creep is the main deformation mechanism. Continuous dynamic recrystallization occurred at 843 K, while discontinuous dynamic recrystallization occurred at 993 K. Comparison of the current results with the previous work on the Cu-4.9Sn alloy indicates that the increase of Sn concentration in Cu matrix extends the regime of solute drag creep and delays the onset of PLB to a lower temperature and a higher strain rate.

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1. Introduction

When copper (Cu) is alloyed with a small amount of tin (Sn), the alloy forms a solid-solution phase (α) that has a good strength compared to pure Cu and other Cu alloys due to the high solid solution strengthening effect caused by the large atomic size difference between Cu and Sn [1]. The Cu-Sn alloys with Sn less than 5 wt.%, have a good plastic working property and it is widely used as electrical connectors [2]. According to the Cu-Sn phase diagram, the solubility of Sn in Cu is as high as ~ 16 wt.% at high temperatures between 773 and 923 K, but it sharply decreases as the temperature approaches room temperature [3,4]. However, due to the low precipitation rate of the ε phase (Cu3Sn) that occurs below 623 K [5], the alloy can retain
the single $\alpha$ phase with a Sn amount up to approximately 10 wt.% at room temperature after an appropriate homogenization treatment. Furthermore, a concentration of Sn near 10 wt.% corresponds to the upper limit for avoiding formation of the hard and brittle $\delta$ (Cu$_4$Sn$_2$) phase during cooling [3,4]. The Cu-10 wt.% Sn alloy shows excellent mechanical properties with a high strength and high ductility as well as high corrosion resistance in salt water; these properties enable application in navigation and machinery industries, such as in marine parts and high-precision bearings [6,7].

Because hot working processes are often required in the production of high-performance mechanical parts, studies on the hot compressive deformation behavior and the optimum hot working conditions of Cu-Sn alloys are important. However, limited works on these aspects are available. Wang and Nix [8] studied the high-temperature tensile creep behavior of a Cu-4.9 wt.% Sn alloy in the temperature range of 726 and 993 K and at low strain rates below $10^{-3}$ s$^{-1}$. They showed that solute drag creep associated with a stress exponent ($n$) of 3, where dislocations glide in a viscous manner due to their interaction with solute atoms, is a rate-controlling deformation mechanism at strain rates below $10^{-4}$ s$^{-1}$ and that the power-law breakdown (PLB) occurs at high strain rates. Hui et al. [9] and Bayle et al. [10] studied the microstructural change after hot deformation in Cu-Sn alloys. Hui et al. [9] examined the microstructures of the Cu-6.5 wt.% Sn alloy after hot compression in the temperature range between 773 and 973 K at a given strain rate of $10^{-2}$ s$^{-1}$. They claimed that the main mechanism of dynamic recrystallization that is responsible for flow softening during compressive deformation was discontinuous dynamic recrystallization (DDRX). Bayle et al. [10] conducted hot compression tests on a Cu-9.2 wt.% Sn alloy at temperatures from room temperature up to 1023 K in the strain rate range of $10^{-3}$ to $10^{-1}$ s$^{-1}$. The results showed that the progress of recrystallization in the Cu-9.2 wt.% Sn alloy was slower and the recrystallized grain size was smaller than those in pure copper.

In the present work, we studied the hot compression properties of a Cu-10 wt.% Sn (Cu-10Sn) alloy at high temperatures (in the range between 843 and 993 K) at different strain rates (in the range between $10^{-3}$ and $10^{1}$ s$^{-1}$). Based on the obtained data, the high-temperature deformation behaviors and mechanisms were analyzed, and the results were compared with those of Cu-4.95Sn alloy [8] and pure Cu [8,11] to understand the effect of addition of Sn on characteristics of hot deformation of Cu. The processing map, which provides information regarding the optimum conditions for hot working for producing defect-free products with homogeneous and fine microstructures [12], was constructed based on an analysis of the flow stress-strain curves as a function of strain rate and temperature at different strain levels. Finally, the microstructures after hot deformation were examined to find the effect of hot deformation parameters on microstructure evolution.

### 2. Experimental procedures

To prepare the Cu-10 wt.% Sn (Cu-10Sn) alloy, pure Cu was first melted at approximately 1423 K in a graphite crucible by using a high-frequency induction melting furnace, and then pure Sn was charged to prepare an alloy melt. The molten Cu alloy was deoxidized by adding a Cu-15%P alloy. The P reacted with the dissolved oxygen present in the molten Cu to form P$_2$O$_5$, which was removed as slag floating on the surface of the molten Cu. An inert gas was bubbled through the molten liquid to remove hydrogen dissolved in the molten alloy. The deoxidated and dehydrogenated liquid was cast into a slab and solidified. The cast alloy ingot was homogenized at a temperature of 1023 K for 10 h and then water cooled. The chemical composition of the Cu-Sn alloy was 10.15Sn-0.005P-0.009Zn-0.009Ni-0.006Mn-0.006Fe-0.005Al-0.01Si-0.007P balance Cu (by wt.%).

To identify the phases in the Cu-10Sn alloy, X-ray diffraction measurements were carried out on a D8 Advance Bruker diffractometer (Germany) with a step size of 0.01° using CuKα radiation in the 2θ range from 30 to 100 degrees.

Uniaxial compression tests (using cylinder-shaped specimens with a diameter of 10 mm and a height of 12 mm) were performed on the Cu-10Sn alloy at temperatures in the range of 843 to 993 K at various strain rates in the range of $10^{-3}$ to $10^{1}$ s$^{-1}$ using a Gleeble 3500 thermomechanical simulator unit. To reduce friction and temperature gradient, a tantalum foil and a nickel-based lubricant were used between the sample and anvil. During the compression tests, each sample was heated to the test temperatures at a heating rate of 10 K/min and held for 5 min before the start of the compression test. The samples were deformed to a total true strain of 1.2 and then taken out of the Gleeble chamber and cooled on a cold copper plate. A K-type thermocouple was spot welded to the middle of the longitudinal length of the sample to monitor the temperature during the compression test. The adiabatic temperature rise was recorded for each compression specimen using the thermocouple. The true stress–true strain curves were plotted based on the load-stroke data obtained from the compression tests. For a given strain and strain rate, the values of flow stress were corrected for the adiabatic temperature rise by plotting the linear interpolation between ln $\sigma$ and 1/T, where $\sigma$ is the flow stress and T is the instantaneous temperature during the test, and then extrapolating back to the initial test temperature [13].

The microstructures after hot deformation were examined by electron back-scattering diffraction (EBSD). The microstructure observation was made at a location that was 1/3 away from the surface toward the center of the compressed sample. For the EBSD measurement in the field-emission scanning electron microscope (S-4300SE, HITACHI, Schaumburg, IL, USA), the samples were polished using a 1 μm diamond paste and then 0.04 μm colloidal silica after the conventional polishing procedures. The EBSD data were analyzed with the aid of the TSL software (Version 7.0), which sets up a tolerance angle of 5° and a confidence index value >0.1. The EBSD step size was 0.2 μm. The grain orientation spread (GOS) method was used to determine the fractions and sizes of dynamically recrystallized (DRX) grains. The grains with GOS value ≤2° were considered DRX grains.

### 3. Results

Fig. 1(a) shows the optical microstructure of the as-received cast alloy. The microstructure comprises an $\alpha$ matrix and
δ(Cu14Sn11) phase, which was confirmed by the XRD result shown in Fig. 1(b). The presence of δ in the as-received microstructure might have resulted from the segregation of Sn at the grain boundaries during the solidification process. The reason for this speculation is because according to the Cu-Sn phase diagram, δ phase does not form in the alloy with a Sn content ≤10 wt.% in the equilibrium state. Fig. 1(c) shows the microstructure after the homogenization treatment conducted at 1023 K for 10 h. The obtained microstructure comprises the α matrix only, which was confirmed by the XRD result shown in Fig. 1(d). The grain size of the homogenized Cu-10Sn alloy was approximately 700–900 μm.

Fig. 2(a)–(e) shows the true stress-true strain curves of the homogenized Cu-10Sn alloy obtained from a series of compression tests performed at different strain rates and temperatures. The solid symbols represent the flow-stress data corrected for the temperature increase during deformation due to adiabatic heating. The adiabatic-temperature increase was only significant at high strain rates above 1 s⁻¹. The following are observed from the stress-strain curves. First, an abrupt decrease in the yield stress occurs at the very beginning of the plastic deformation under some experimental conditions. As temperature increases, this yield-drop like behavior becomes more pronounced and the upper limit of strain rate to display this phenomenon increases. This may be because the contribution of solute drag creep to total deformation and the upper limit of the strain rate for solute drag creep increases as the temperature increases, which will be shown later. The yield-drop like phenomenon at elevated temperatures has been reported in the class I type solid solution alloys such as Al-Mg and Al-Cu [14], which exhibit solute drag creep as a rate-controlling deformation mechanism in the specific temperature and strain-rate range. According to dynamic deformation model [14], discontinuous yielding (yield-drop like behavior) occurs as a result of rapid generation of high density of new mobile dislocations upon initial plastic straining. Second, after an abrupt decrease in the yield stress occurs, the flow stress increases by strain hardening, reaches a maximum value, decreases with an increase in strain and then a steady-state-like behavior appears.

The empirical relationship among strain rate, temperature and flow stress in pure metals and metallic alloys can be described by the sine hyperbolic Garofalo equation that considers the dependence of the steady-state strain rate on the flow stress and temperature for both high and low stresses [15]:

\[
i = A \sinh(\alpha \sigma)^n \exp\left(\frac{Q_c}{RT}\right)
\]

where \(i\) is the strain rate, \(Q_c\) is the activation energy for creep, \(A\) is the material constant, \(R\) is the gas constant, \(n\) is the stress exponent, and \(\alpha\) is the fitting parameter. When \(\alpha \sigma < 0.8\) (at low stresses), Eq. (1) can be reduced to Eq. (2), which represents power law creep, and when \(\alpha \sigma > 1.2\) (at high stresses),
Eq. (1) can be simplified into Eq. (3), which represents power-law breakdown (PLB):

\[ \dot{\varepsilon} = A\sigma^{n_1}\exp\left(-\frac{Q_c}{RT}\right) \]  

\[ \dot{\varepsilon} = A\exp(\dot{\varepsilon}_0)\exp\left(-\frac{Q_c}{RT}\right) \]

Where \( \dot{\varepsilon} = a_n_1 \). Ideally, \( n_1 = n \). In this work, the values of \( a \) and \( Q_c \) were calculated assuming that they are not a function of strain rate, temperature and strain in the steady state. Justification for this assumption is presented in the reference of 16. To evaluate \( a \), the experimental data at high stresses (for determination of \( \beta \) and low stresses (for determination of \( n_1 \)) were used separately.
Fig. 3(a) shows the plot of $\ln \dot{\epsilon} - \ln \sigma$ for the Cu10Sn alloy at $\varepsilon = 0.6$ for the measurement of $n_1$ values at different temperatures from the linear fit to the data at low strain rates where a linear relationship is observed. The $n_1$ value measured in the temperature between 843 and 993 K gradually decreases as temperature increases, and the values are in the range of 3.6 and 4.6. The average $n_1$ value is 3.9. The measured $n_1$ values are larger than the typical value associated with solute drag creep ($n_1 = 3$) but smaller than the typical value associated with lattice-diffusion controlled dislocation climb creep ($n_1 = 5$). Fig. 3(b) shows the plot of $\ln \dot{\epsilon} - \sigma$ for the determination of $\beta$ at different temperatures at $\varepsilon = 0.6$. From the linear fitting of data points for the strain rates above $10^{-1}$ s$^{-1}$, the $\beta$ values were determined and are in the range of 0.036–0.038. The average $\beta$ value is 0.038. The $\alpha$ value calculated using $\alpha = \beta/n_1$ is in the range of 0.0082–0.01, tending to increase gradually as temperature increases. The average $\alpha$ value is 0.0091. This value was used in calculating the $Q_c$ value in Eq. (1).

The $Q_c$ value in Eq. (1) can be calculated using Eq. (4), which is obtained by differentiating Eq. (1) with respect to the strain rate and temperature:

$$Q_c = R \left[ \frac{\partial \ln[\sinh(\alpha \sigma)]}{\partial (1/T)} \right]_{\dot{\epsilon}} \left[ \frac{\partial \ln \dot{\epsilon}}{\partial \ln[\sinh(\alpha \sigma)]} \right] - \ln S$$  \hspace{1cm} (4)

where $n$ is determined by measuring the average slopes of the curve fits in the plot of $\ln \dot{\epsilon} - \ln \sinh(\alpha \sigma)$ at different temperatures (Fig. 3(c)) and $S$ is determined by measuring the average slopes of the curve fits in the plot of $\ln \sinh(\alpha \sigma)$ - $1/T$ at different strain rates (Fig. 3(d)). The calculated $Q_c$ value is 171.3 kJ/mol. The plot of $\ln Z - \ln \sinh(\alpha \sigma)$ at $\varepsilon = 0.6$ (where $Z$ is the Zener–Hollomon parameter, $Z = \dot{\epsilon} \exp(\Delta H/RT)$), where $\alpha = 0.0091$ MPa$^{-1}$ and $Q_c = 171.3$ kJ/mol, are shown in Fig. 4. A good linear fit is obtained ($R^2 = 0.98$). The $n$ value calculated from the slope of the curve is 3.3, supporting that solute drag creep is the rate controlling deformation mechanism in the power-law creep regime. The $Q_c$ value for the power law creep regime can be directly calculated by differentiating Eq. (2) with respect to temperature under a given flow stress:

$$Q_c = -R \left[ \frac{\partial \ln \dot{\epsilon}}{\partial (1/T)} \right]_{\sigma}$$  \hspace{1cm} (5)

The average of the $Q_c$ values measured from the plot of $\ln \dot{\epsilon} - 1/T$ at $\varepsilon = 0.6$ at different flow-stress levels (Fig. 5) using Eq. (5) is 170.1 kJ/mol. This value is very similar to that obtained using Eq. (4), indicating that the $Q_c$ values for the power-law creep and PLB are similar. The $Q_c$ value obtained from Cu10Sn alloy, 170.1–171.3 kJ/mol, is smaller than the activation energy for self-diffusion in pure Cu (197 kJ/mol [17]). During solute drag creep, it is reasonable to consider that the creep rate is controlled by chemical interdiffusivity ($-D$) between the solute and solvent atoms because the solute atmosphere diffuses with the dislocations by exchanging positions with solvent atoms in the dislocation path. When the concentration of solutes is dilute, $-D = D_{\text{sol}}$ where $D_{\text{sol}}$ is the intrinsic diffusivity of solutes with the activation energy of $Q_{\text{sol}}$.

According to Barrett et al. [18], the $Q_c$ value determined at a constant stress can be higher than the activation energy for atomic diffusion when the temperature dependence of elastic modulus is important. When this modulus effect is considered, the $Q_c$ value at a constant stress is given by:

$$Q_c = Q_{\text{sol}} - n \frac{T^2}{C} \frac{dG}{dT}$$  \hspace{1cm} (6)

where $G$ is the shear modulus (≈47129.7–16.8T MPa for pure Cu [17]). The $Q_{\text{sol}}$ value calculated from Eq. (6) (with $Q_c=170.1$ kJ/mol and $dG/dT=-16.8$ MPa) is 159 kJ/mol. This value is very close to the activation energy for $D_{\text{sol}}$ in Cu-Sn alloys (≈156 kJ/mol [19]).

A processing map comprises a power dissipation map and an instability map, which represent the characteristics of power dissipation by microstructural evolution and the flow instability regions, respectively, under different strain rate-temperature conditions [20]. The efficiency of the power
dissipation, $\eta$, is defined to represent how efficiently the power is consumed by a microstructure change during plastic deformation [20], and the power dissipation map is plotted using the $\eta$ values. The $\eta$ values were calculated using the strain-rate sensitivity exponent ($m = 1/n_1$) that was determined based on fits to the log $\sigma$ vs. log $\dot{\varepsilon}$ curves obtained from the true stress-true strain curves as a function of strain rate, temperature and strain using a cubic spline:

$$\eta = \frac{2m}{m + 1}$$

Prasad’s criterion [20] has been popularly used in constructing the flow instability map, and according to it, the instability parameter $\xi$ is described as follows:

$$\xi = \frac{\partial \ln (m/(m + 1))}{\partial \ln \dot{\varepsilon}} + m < 0$$

When $\xi$ becomes negative, deformation in the material is predicted to be unstable. In the instability domain, flow localization or cracking can occur, leading to the formation of defect-containing inhomogeneous microstructures in the products after hot working. Therefore, in designing a hot working process, this regime should be avoided.

The power dissipation maps for the Cu-10Sn alloy were constructed using the $\eta$ values as a function of the temperature and strain rate at different strain levels (at $\varepsilon = 0.2, 0.4, 0.6$ and 0.8), and the results are shown in Fig. 6(a)–(d). The contour values in the power dissipation maps represent the $\eta$ values (by percentage) and the shaded areas represent the flow instability regions. The $\eta$ values are plotted as a function of temperature and strain rate at different strains at 843 and 993 K.

**Fig. 6** – The processing maps for the Cu-10Sn alloy at (a) $\varepsilon = 0.2$, (b) 0.4, (c) 0.6 and (d) 0.8.

**Fig. 7** – The $\eta$ values of the Cu-10Sn alloy and the pure Cu [11] plotted as a function of strain rate at different temperatures at different strains at 843 and 993 K.
of strain rate at different temperatures at different strains at two temperatures of 843 and 993 K (Fig. 7). There are several important findings from Figs. 6 and 7. First, the flow instability region appears at temperatures less than 933 K at high strain rates above $10^{-2} - 5 \times 10^{-1}$ s$^{-1}$ at all strain levels. Above 933 K, however, the flow instability region does not exist even at the highest strain rate of 10 s$^{-1}$. As the strain increases, the instability region shrinks. The boundary of the instability region is pushed to an increasing strain rate. No further change in the instability region occurs beyond $\varepsilon = 0.6$. Second, the $\eta$ value is 37–40% at low strain rates at all temperatures. These values decrease with increasing strain rate because the $m$ value decreases with increasing strain rate. The rate that the $\eta$ value decreases, however, is lower at a higher temperature. For example, at the highest strain rate of 10 s$^{-1}$, the $\eta$ value is 31–35% at 993 K, while it is only 3–8% at 843 K. This results because the upper limit of the strain rate for solute drag creep increases as the temperature increases. Third, the $\eta$ value increases as the strain level increases from 0.2 to 0.4 at all the testing conditions but becomes saturated above 0.4. Fourth, the $\eta$ values of the Cu-10Sn alloy are considerably larger than those of the pure Cu [11], except at the high strain rates at 843 K.

The microstructures after the compressive deformation were examined by optical microscopy (at a low magnification) and EBSD (at a high magnification) at 843 and 993 K at the two strain rates of $10^{-3}$ and 10 s$^{-1}$ (Fig. 8(a)–(h)). The fraction of the dynamically recrystallized (DRX) grains, the size of the DRX grains, the average grain size, and the fractions of twin boundaries are plotted in Fig. 9(a)–(d), respectively. The following are

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**Fig. 8** – The optical micrographs and the GB/IPF maps of the samples after the compressive deformation (a and b) 843 – $10^{-3}$ s$^{-1}$, (c and d) 843 K – 10 s$^{-1}$, (e and f) 993 K – $10^{-3}$ s$^{-1}$ and (g and h) 993 K – 10 s$^{-1}$. In the EBSD maps, low-angle grain boundaries ($2^\circ \leq \theta < 5^\circ$), intermediate-angle grain boundaries ($5^\circ \leq \theta < 15^\circ$) and high-angle grain boundaries ($15^\circ \leq \theta$), are represented by green, blue and red colors, respectively. The insets show the inverse pole figures.
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Fig. 9 – (a) The fraction of DRXed grains, (b) the size of DRXed grains, (c) the average grain size, and (d) the fractions of twin boundaries.

Fig. 10 – The GB maps of the sample deformed at (a) 843 K – 10^{-3} \text{s}^{-1} and (b) 993 K – 10^{-3} \text{s}^{-1}. In the EBSD maps, low-angle grain boundaries (2° \leq \theta < 5°), intermediate-angle grain boundaries (5° \leq \theta < 15°) and high-angle grain boundaries (15° \leq \theta), are represented by light blue, blue and black colors, respectively. Different colors were used to show HABs with a coincidence site lattice (CSL) relationship. The Σ3 twin boundaries are shown as red lines, while the Σ9 and Σ27 twin boundaries are displayed as yellow lines and green lines, respectively.

noted from Figs. 8 and 9. First, inhomogeneous microstructures containing deformed grains and recrystallized grains are observed at 843 K at the two strain rates, whereas homogeneous recrystallized microstructures are observed at 993 K at the two strain rates. The fractions of DRX grains are 0.92 and 0.83 at strain rates of 10^{-3} and 10 \text{s}^{-1}, respectively, at 993 K, which are higher than those (0.61 and 0.58) at 843 K. The size of the DRX grain sizes at 10^{-3} and 10 \text{s}^{-1} at 843 K are 10.6 and 7.7 \text{μm}, respectively, which are smaller than those measured at 993 K (66.9 and 20.0 \text{μm}). These results indicate that at a higher temperature and at a lower strain rate, a higher fraction of DRX grains with a larger size are obtained, agreeing with a typical DRX behavior in metals. A larger fraction of recrystallized grains at a higher temperature is due to the effect of the temperature on the DRX kinetics and a larger DRX grain size at a higher temperature is due to the increased grain boundary
mobility at high temperatures. The decrease in the fraction and size of the recrystallized grains at an increased strain rate is due to the decreased time for the formation of DRX nuclei. Second, the fraction of twin boundaries is higher at 993 K than at 843 K (0.12–0.21 vs. 0.05–0.07). From the magnified GB maps of the sample deformed at 843 K – 10^{-3} s^{-1} and 993 K – 10^{-3} s^{-1} (Fig. 10(a) and (b)), it is recognized that small DRX grains are free of twins, while many of large DRX grains contain twins. This observation indicates that twins generate during the growth of the recrystallized grains. This type twins can be regarded as annealing twins owing to the absence of plastic deformation in the recrystallized structure. Increased temperature will facilitate the formation of twin boundaries because this factor increases the growth of recrystallized grains. Primary twins Σ3, characterized by a 60° misorientation around the (111) plane are most abundant, and higher order twins, including Σ9 (38.9° / 101) [21], Σ27a (31.6° / 110) [22] and Σ27b (35.4° / 210) [23], are much less frequently observed. Twins were rarely observed in the pure Cu after compressive deformation under the same experimental conditions, where fractions of DRX grains were low (less than 0.2) [11]. Third, the ND/⟨101⟩ texture developed after hot compressive deformation. It is known that ⟨110⟩ fiber texture is the characteristic for compression texture of FCC copper [24] and the rotation of grains toward ⟨110⟩ caused by dislocation slip leads to the formation of ⟨110⟩ fiber texture [25]. The maximum texture intensities are 3.0 and 3.4 mrд at 843 K, which are weaker (1.5 and 2.5 mrд) than those at 993 K. This is because of the occurrence of a higher degree of DRX at 993 K.

Generally, discontinuous dynamic recrystallization (DDRX) occurs in cubic metals with a low or medium stacking fault energy (SFE) [26], and it features nucleation by grain boundary bulging followed by growth of the recrystallized grains. Cu has a medium SFE, but it has been reported that the addition of Sn to Cu decreases the SFE of Cu [27]. Unlike in DDRX, continuous dynamic recrystallization (CDRX) occurs by the generation of low-angle grain boundaries, which gradually transform to high-angle grain boundaries with increasing deformation [28]. To clarify the type of DRX that operates during compressive deformation, misorientation analysis was performed along the vector lines within coarse grains shown in Fig. 8(a)–(d). The point-to-point (local) and point-to-origin (cumulative) misorientations were plotted and the results are shown in Fig. 11(a)–(d). At 843 K, cumulative orientation gradients frequently exceed 15°, while the local misorientation does not exceed 5–6° (Fig. 11(a)–(c)). CDRX is characterized by a progressive increase in the misorientation angle from low to high values of 10–15° upon straining. Therefore, the observation of the point-to-origin misorientation that often exceeds 10–15° and the presence of many low-angle grain boundaries within grain interiors at 843 K indicate that recrystallization occurs by CDRX. Furthermore, incomplete discontinuous intermediate- and high-angle grain boundary segments are frequently observed within grains in the GB maps, which...
are connected by low-angle grain boundaries, as shown in Fig. 10(a). This observation suggests that discontinuous intermediate- and high-angle grain boundaries have evolved from low-angle grain boundaries by the continuous absorption of dislocations during deformation. At 993 K (Fig. 11(d)), the local and cumulative misorientations do not exceed values of 2 and 4°, respectively. Low cumulative misorientations and local misorientations along the grain boundaries and grain interior suggests that recrystallization occurs by DDRX.

4. Discussion

Fig. 12 shows the plot of the log Z vs. log(σ/G) for the Cu-4.95Sn alloy from the creep data in tension [8] and for the Cu-10Sn alloy at ε = 0.6 constructed using Q_{sol} = 156 kJ/mol. The Cu-4.95Sn alloy exhibits n_1 = 3.5 at small Z values and the onset of PLB at ln Z = 6.8, while the Cu-10Sn alloy exhibits n_1 = 3.8 at small Z values and the onset of PLB at ln Z = 8.5. The two data sets of the Cu-4.95Sn and Cu-10Sn alloys are close to each other and macroscopically well correlated with a single curve. From this result, it can be speculated that at small Z values, solute drag creep is the dominant deformation mechanism in the Cu-Sn alloys but the deformation mechanism changes from solute drag creep to dislocation climb creep and to PLB as Z value increases. The Al-Mg [29] and In-Sn alloys [30] exhibit the similar transitions with increasing strain rate from n_1 = 3 to n_1 = 5 and to PLB.

For solute drag creep, creep deformation is assumed to be limited by glide process of dislocations dragging solute atmospheres and Weertman proposed the equation for the steady state creep rate for this case [31]:

\[ \dot{\epsilon} = \frac{1}{3e^2C_o} \left( \frac{kT}{G\delta^3} \right)^2 \left( \frac{D\delta}{kT} \right) \left( \frac{\sigma}{G} \right)^3 \] (9)

where ε is the solute-solvent size factor, C_o is the solute concentration, k is Boltzmann’s constant (=1.38 x 10^{-23} J/K), b is the Burgers vector, −D can be reduced D_{sol} when the solute concentration is low. With b = 2.56 x 10^{-10} m for pure Cu [16], ε = 0.2241 [32], C_o = 0.027 or 0.056 and D_{sol} = D_{Sn} = D_e exp(−156000/kT) m^2 s^{-1} where D_e is a function of Sn concentration: D_e = 2 x 10^{-3} x 10^{-3.3C_o} m^2 s^{-1} [33], the strain rates of the Cu-4.95Sn and Cu-10Sn alloys in Eq. (9) were calculated as a function of (z). The prediction by Eq. (9) reasonably well agrees with the experimental data, as shown in Fig. 12.

The increase of C_o of Sn is expected to decrease the strain rate, but it also increases D_{sol}. As the effect of the latter is more significant, the strain rates of Cu-10Sn alloy are predicted to be slightly higher than those of Cu-4.95Sn alloy, agreeing with the experimental observation.

The delay of the onset of PLB with increase in the amount of added Sn may be explained in terms of a breakaway-stress concept assuming that dislocations break away from the solute atmosphere at a critical breaking stress (σ_{break}) [34]:

\[ \sigma_{break} = \frac{W_m^2 C_o}{5b^4kT} \] (10)

where W_m is the binding energy between the solute atom and the dislocation (−1/2σ[(1 + v)/(1 − v)]G[A_{V_{Cu}}] [34], where v is the Poisson’s ratio for pure Cu (0.33 [16]) and A_{V_{Cu}} is the difference in volume between solute (Sn) and solvent (Cu) atoms (9.85 x 10^{-20} m^3 [32]). The values of W_{break} calculated by Eq. (10) at 843 K are 8.85 x 10^{-3} and 1.84 x 10^{-2} for atomic Sn concentrations of 0.027 and 0.056, respectively. These calculated values are in agreement (within a factor of 2.4–3.0) with the experimental values of z for the onset of PLB (indicated by arrows in Fig. 12) and the ratio of W_{break} for the Cu-10Sn to Cu-4.95Sn alloys (2.1) is close to the ratio of z associated with PLB for the Cu-10Sn to Cu-4.95Sn alloys (1.7), suggesting that the delay of PLB is related to the critical conditions for breakaway of dislocations from their solute atmospheres.

Fig. 13(a) shows the plot of the log Z vs. log(σ/G) for the pure Cu [8,11] and the Cu-Sn alloys. In plotting, the activation energy for self-diffusion in Cu (197 kJ/mol) and the activation energy for Sn diffusion in Cu-Sn alloy (156 kJ/mol) were used for the pure Cu and the Cu-Sn alloys, respectively. It is obvious that the n_1 value of the Cu-Sn alloys (3.5) is smaller than that (4.8) of the pure Cu at small Z values. For the pure Cu, lattice-diffusion controlled dislocation climb creep is a governing deformation mechanism. Fig. 13(b) shows the plots of the log Z vs. log(σ/G) for the pure Cu and the Cu-Sn alloys constructed using the average (176.5 kJ/mol) of the two Q values of the pure Cu and the Cu-Sn alloys. The use of the same Q value for plotting of the data of the two materials, which is possible when their Q values are not so much different, allows for a direct comparison of flow-stress and n_1 values at any given temperature and strain rate. The onset of PLB occurs in the pure Cu at a Z value, which is notably smaller than the Z values for the Cu-Sn alloys. This result indicates that as Sn is added to Cu, the onset of PLB is delayed to a lower temperature and a higher strain rate. The increase of n_1 value and the delay of the onset of PLB due to the solute drag effect on dislocation glide mobility explain why the n_1 values of the Cu-10Sn alloy are larger than those of the pure Cu in the wide range of tem-
temperature and strain rate (Fig. 7). Flow stresses of the Cu-4.9Sn and Cu-10Sn alloys are slightly lower than those of the pure Cu at small Z values, but higher at large Z values.

Many studies [35–38] have shown that DRX grain size and the fraction of DRX grains are well correlated into a single curve with the Zener-Holloman parameter: DRX fraction and grain size decrease with increasing Z. The plots in Fig. 9(a) and (b), however, show that the DRX parameters are poorly correlated (as shown by dotted lines) with the Zener–Holloman parameter. This may be because different DRX mechanisms operate at 843 and 993 K, as discussed in Fig. 11. CDRX is known as a one-stage process, without a long-range migration of high-angle grain boundaries [28,39]. In CDRX, subgrains evolve into new DRX grains progressively such that they can hardly grow. In contrast, DDRX is a two-stage process comprising the nucleation of new grains at the grain boundaries of initial grains and long-range migration of high-angle grain boundaries of the newly formed grains by consuming the deformed grains [28,40]. Thus, a finer DRX grain size is expected to form during CDRX. This explains why the DRX grain sizes at 843 K are smaller than those at 993 K (Fig. 9(b)). On the other hand, the kinetics of DDRX is slower than the kinetics of CDRX [28]. This is because transformation from low-angle grain boundaries into high-angle grain boundaries by DDRX process requires a large strain [28]. This explains why the fractions of DRX grains at 993 K are higher than those at 843 K (Fig. 9(a)).

The Cu-10Sn alloy exhibits considerably higher efficiencies of power dissipation compared to the pure Cu. This result agrees with the microstructural observations because the pure Cu exhibits the low fractions of DRX grains (0.1–0.2) and high angle grain boundaries (0.1–0.3) [11], while the Cu-10Sn alloy exhibits the high fractions of DRX grains (0.6–0.9) and high angle grain boundaries (0.7–0.9).

![Diagram of log Z vs. log(σ/G)](image)

**Fig. 13** (a) The plot of the log Z (with Q = 197 kJ/mol for pure Cu and Q = 156 kJ/mol for Cu-Sn alloys) vs. log(σ/G) for the pure Cu and Cu-Sn alloys. (b) The plot of the log Z (with Q = 176.5 kJ/mol for pure Cu and the Cu-Sn alloys) vs. log(σ/G) for the pure Cu and the Cu-Sn alloys. The plot of the creep data of the pure Cu, which had been plotted with Q = 197 kJ/mol in the reference of 8, were replotted using Q = 176.5 kJ/mol. Arrows indicate the onsets of PLB (n1 ≥ 7).

5. **Conclusions**

Hot compression tests were performed on the Cu-10Sn solid solution alloy in the temperature range of 843–993 K and in the strain rate range of 10⁻³–10⁻¹ s⁻¹, and the following results were obtained.

1. A yield-drop like phenomenon was observed in the true stress–true strain curves, indicating a strong interaction between the Sn solutes and mobile dislocations.
2. The activation energy for plastic flow (159 kJ/mol) was reasonably close to that for the intrinsic diffusivity of Sn in Cu (156 kJ/mol). The stress exponent measured at low strain rates decreased from 4.6 to 3.6 as the temperature increased from 843 to 993 K, suggesting that solute drag creep becomes important as the temperature increases.
3. DDRX is the dominant dynamic recrystallization mode at 993 K, but as the temperature decreases, the contribution from CDRX becomes more significant.
4. The processing maps indicate that the Cu-10Sn alloy has good hot workability at temperatures above 912 K: no flow instability appears up to a very high strain rate of 10⁻¹ s⁻¹ and the power dissipation efficiency is high. This results from an increase in the contribution of solute-drag creep to the total deformation and the extension of the solute-drag creep dominant regime to an increased strain rate with increasing temperature.
5. The addition of the large amount of Sn to Cu is beneficial because the upper limit of the strain rate (temperature) for solute drag creep increases (decreases), leading to the increase of the efficiencies of power dissipation and the delay of the onset of PLB to a lower temperature and a higher strain rate.

**Conflict of interest**

The paper submitted has no conflict of interest.

**Data availability**

The raw/processed data required to reproduce these findings cannot be shared at this time as the data also forms part of an ongoing study.
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