Hot tensile deformation study and constitutive modeling of new high zinc containing hot rolled Al–Zn–Mg–Cu alloy

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Keywords: Al alloy, hot rolling, hot tensile deformation, constitutive model, microstructural analysis, deformation mechanism

Abstract
Hot tensile deformation behavior of a new kind of hot rolled high zinc containing Al–9.6%Zn–3.1% Mg–1.84%C–Cu alloy was investigated under different deformation temperatures (250 °C–400 °C) and strain rates (0.000 05–0.000 05 s⁻¹) conditions. From the true stress-true strain diagrams, it is observed that peak stress decreases with increasing temperature and decreasing strain rate. Exponent type equations following Arrhenius model were used to predict peak stress at the experimental conditions and the values were compared with the experimental values. Zener–Hollomon parameter (Z) has been determined at various test conditions and activation energy has been found to be 146.2 kJ mol⁻¹ which is slightly greater than the bulk self-diffusion energy of pure Al (142 kJ mol⁻¹) and consistent with the previous findings. Microstructural analysis of the deformed samples shows that deformation mechanism transforms to dynamic recrystallization from dynamic recovery with the decrease of the Z value. Additional XRD, SEM, DSC and EDS analysis were performed to determine the type of the precipitates present in the sample before deformation and analyze their effect on the deformation mechanism.

Introduction
Flow stress prediction of a hot deformed material is a convoluted task as many metallurgical phenomena such as work hardening, dynamic softening, and dynamic recrystallization simultaneously occur in many materials. It is highly dependent on the change in temperature and strain rate. So, it is necessary to understand the relationship among flow stress, temperature, strain rate and microstructures to optimize process parameters. Generally, hot tensile and hot compressive tests are carried out to predict flow behavior. Recently, many researchers have conducted hot tensile tests to study the deformation behavior, kinetics and fracture characteristics on different alloys systems such as Al–Cu–Mg–Zn alloy [1], Al–Cu–Zn alloy [2], Pure Ti [3], Mg/space alloys [4–6], Boron steel [7] etc. Various models have been proposed based on the results to predict flow behavior. In all those models, some equations are proposed which estimates the power or load required for the hot deformation process and flow behavior using true stress strain curve [8–16]. Some researchers have also made effort to simulate various test results. As long as the equations are reliable, simulation can also produce valid results. That is why, serious efforts have been made to construct appropriate model for the flow behavior prediction. Arrhenius model has been proved best among all the available models as various effect of the change of deformation temperature and strain rate is considered here [17–19].

Al–Zn–Mg–Cu is one of the important alloys in the aerospace and structural industries due to its excellent combination of properties such as high strength to weight ratios, good stress corrosion cracking resistance, and toughness [20–23]. Comparatively high amount of alloying element provides the alloy with high room temperature strength, but this property highly deteriorates at high temperatures. So, flow behavior modeling is very important for these types of quaternary systems to circumvent formation of defect and optimize thermo-mechanical forming processes such as rolling, forging, extrusion etc. Zhou et al [1] studied the hot tensile deformation behavior of Al–Zn–Mg–Cu alloy and provided a constitutive model. Hot compression deformation behavior of AA7085 alloy [24], 7015 Al alloy, as-quenched 7005 Al alloy [25], as cast and
homogenized Al 2024 alloy [26], Al–Zn–Mg–Cu–Zr alloy [27] were investigated by various researchers at elevated temperatures and strain rates. To explain the hot compression deformation behavior of Al–Zn–Mg–Cu alloy, Lin et al [28] developed a new type of phenomenological constitutive model where material constants are expressed as functions of forming temperature, strain and strain rate. They optimized the hot deformation processing parameters of 7075 aluminum alloy system by designing processing map. Quan et al [29] developed an improved Arrhenius model for the as-extruded 7075Al alloy to characterize the flow behavior with variable parameter. A large number of efforts have been made till now to study the hot deformation behavior of Al–Zn–Mg–Cu alloys using hot compression tests. However, little attention was paid on the hot tensile deformation behaviors and microstructural study of the same system. Informations that are obtained from compression tests can also be obtained from the tensile tests. Nonetheless, tensile test provides some additional information such as the ductility, upper and lower yield point, fracture strength and behavior etc of the material. For metallic materials, these additional informations are also very important.

In this research, a new kind of high Zinc containing AL–Zn–Mg–Cu alloy has been produced and the effects of various thermo-mechanical parameters on the hot tensile deformation behavior and microstructural changes of this Al–Zn–Mg–Cu alloy have been investigated. Hyperbolic sine equations revealed the relationship among peak stress, deformation temperature and strain rate and experimental values show that the equations have good predictability with ±5% relative error. Microstructural evolution study was done to see the variation in softening mechanism in various working conditions. Several characterization techniques were used to understand the structure and relate it to its hot deformation characteristics.

**Experimental**

The alloy was prepared by taking pure Al (99.8%), pure Mg (99.9%), pure zinc (99%) ingots and electrolytic Cu (99.9%), and melting them in a pit furnace. The cast product was obtained by pouring the liquid into a metal mold (pre-heated to 200 °C) followed by air cooling. Composition of the cast product was examined using x-ray fluorescence spectrometer (Shimadzu, Lab center XRF-1800). Chemical composition is given in table 1. The product is then homogenized in the PLF 110/30 Protherm furnace at 420 °C temperature for 6 h. To inspect the phases by obtaining chemical composition, EDS analysis has been performed during acquiring images in scanning electron microscope (FESEM; JEOL JSM 7600F), XRD (PANalytical Empyrean x-ray Diffractometer system) and DSC (Setaram, DSC 131 EVO) analysis have also been done to determine the phases present in the structure.

Homogenized product was then hot rolled and reduction in thickness was about 72% in a 250 × 300 mm two-hi rolling mill. In order to develop the constitutive relations among peak stress, temperature and strain rate, hot uniaxial tensile testing was carried out by computer controlled hot tensile machine (Testometrie M500-50 computer controlled universal materials testing machine) at temperatures of 250 °C, 300 °C, 350 °C and 400 °C and at strain rates of 0.000 05, 0.000 15 and 0.0005 s⁻¹. Tensile specimens were heated to the elevated temperatures at the heating rate of 5 °C min⁻¹ and held at that temperature for 15 min for obtaining uniform temperature. Deformed microstructures were obtained from sections parallel to the tensile axis. Microstructures of the as cast and homogenized, rolled and deformed product were observed in Optica B-600 MET trinocular upright metallurgical microscope after etching with Keller’s reagent and images of same resolution were acquired using OpticaTM Vision Pro software package.

**Results and discussion**

**Microstructure**

Microstructures of the as-cast and homogenized sample are shown in figure 1. In the as-cast structure, a lot of grain boundary segregation and inhomogeneities can be seen. Intermetallics are coarse in size and they are also uniformly distributed along the grain boundaries. Non uniformity is present at a greater volume in the as-cast sample. It is known that segregation at grain boundary has adverse effect on the properties. So, it is necessary to remove segregations. When temperature is increased, diffusion coefficient increases by the following change [30].

| Analyte | Al | Zn | Mg | Cu | Si | Fe |
|---------|----|----|----|----|----|----|
| Wt% amount | 84.83 | 9.6 | 3.1 | 1.84 | 0.31 | 0.19 |
According to this equation, the higher the temperature, higher is the diffusion coefficient. So, when the cast product is treated at a higher temperature, segregation and inhomogeneities are minimized due to diffusion. So after heat treating at 420 °C, grain becomes more uniform and segregation is minimized. Still the homogenized structure contains eutectic grain boundary phases and dispersed phases.

SEM and EDS analysis of the as-cast and homogenized samples were carried out in the backscattered mode (figures 2 and 3). EDS analysis gives an idea about the phases present in the sample. For the as-cast sample, chemical composition at point A shows that the phase contains only Al, Zn, Mg and Cu. This phase can be just Al–Cu–Mg–Zn phase. Segregation is obvious at the grain boundary. In fact, it is highest at the grain boundary. From the grain boundary to the inside of the grains, concentration of the element have decreasing percentages. EDS result is shown in table 2 below.

SEM and optical image of the homogenized sample show that the volume of the interdendritic phases decreases compared to the as-cast sample. These phases also lose their continuity at the grain boundary.

According to EDS result (shown in table 3), Al–Cu–Mg–Zn phase is absent in the homogenized sample. Grain boundary intermetallic phase contain Al₂CuMg, Al₂Mg₂Zn₃. Besides that, matrix contains small MgZn₂, Mg₆Si, Al₃Fe phase precipitates.

Homogenized sample was reduced in thickness from 9 to 2.5 mm using multiple passes by hot rolling at 400 °C, and sample was then prepared for hot tensile testing. Hot rolled microstructure is shown in figure 4. It is seen that, non-uniform grain structure has been produced. Partially equiaxed grains are present at the sample edge whereas central region contains highly deformed, elongated grains aligned in the direction of rolling. This kind of non-homogeneous structure occurs when deformation condition and processing parameters are not
Table 2. Chemical composition of the phases in figure 2.

| Location | Al  | Mg  | Zn  | Cu  | Fe  | Si  | Identification       |
|----------|-----|-----|-----|-----|-----|-----|----------------------|
| A        | 32.56 | 28.14 | 23.71 | 15.59 | 0   | 0   | Al–Cu–Mg–Zn          |
| B        | 34.02 | 26.70 | 23.64 | 15.43 | 0   | 0   | Al–Cu–Mg–Zn          |

Table 3. Chemical composition of the phases in figure 3.

| Location | Al  | Mg  | Zn  | Cu  | Fe  | Si  | Identification       |
|----------|-----|-----|-----|-----|-----|-----|----------------------|
| A        | 47.16 | 20.67 | 19.38 | 12.19 | 0.04 | 0.41 | Al₂CuMg               |
| B        | 23.24 | 34   | 30.27 | 12.49 | 0   | 0   | Al₂Mg₂Zn₃             |
| C        | 68.23 | 9.66 | 16.47 | 5.28  | 0.23 | 0.12 | MgZn₂                 |
| D        | 31.73 | 22.28 | 0.52  | 15.10 | 0.04 | 30.19 | Mg₆Si                |
| E        | 77.17 | 0.94 | 1.25  | 3.44  | 16.09 | 0.43 | Al₃Fe                 |

Figure 3. SEM image of the homogenized sample in the backscattered electron mode.

Figure 4. Microstructure of the hot rolled sample of the studied alloy.
uniform throughout the rolling process. Again the hot rolling was carried out at 400 °C which was not apparently sufficient for the complete recrystallization of the sample.

This kind of microstructure points out that continuous recovery and recrystallization has occurred in the structure. They have occurred to produce defect free regions in the grain to reduce the energy of the system. During rolling, an alloy is deformed in the condition of plane strain compression. This deformation produces ‘pancake’ type lamellar structure by the elongation of grain. During stretching in one direction which is rolling
direction also creates contraction in the perpendicular direction causing grain contraction. As a result, grain boundaries are pushed together and finally adjacent grain boundaries collapse forming new grains. These newly formed grains can have equiaxed, elongated or both shapes depending on the mobility of the grain boundary.

Rolled structure shows that the sample has undergone more recovery than recrystallization. Recovery process usually occurs by dislocation climb and cross slip process at elevated temperature by vacancy migration induced dislocation process [31]. Initial hot deformation of the sample produces dislocations that accumulates in tangles forming sub-boundaries by dynamic recovery process [32]. These sub-grain structures can be stable if deformation condition remains at constant strain rate, stress and temperature. During rolling, all the conditions remained stable for having stable sub-grain structure. Here, stacking fault energy (SFE) here played the most important role in dynamic recovery. Higher the SFE, more the alloy is favored for dynamic recovery by the process of dislocation climb or cross slip. When SFE is lowered, recovery by dislocation faces obstacles. Al alloys have generally high SFE which helped the rolled structure to undergo more recovery than recrystallization.

Recrystallization process actually occurs in the deformed regions by the movement of high angle grain boundaries. Initiation of recrystallization depends on the strain level during hot rolling and critical strain rate. High angle grain boundaries are mainly involved in recrystallization because these boundaries are usually associated with low activation energy and high internal energy required for dislocation movement. Most other regions have low energy and dispersoids are also present throughout the matrix. Dispersoids usually create obstacles in the grain boundary migration process. They can also pin the cell wall, dislocations and grain boundary movement during recrystallization. As a result, they restrict or hinder the movements of dislocations in the deformed grains to produce a fully recrystallized grain. Precipitate size, volume fraction of precipitates and interprecipitate distance mainly plays the most important role to produce significant effect in affecting recrystallization phenomena [33, 34]. In the microstructure, large amount of elongated grains indicate that the size and amount of dispersoids present after hot rolling was numerous. XRD analysis also proved the presence of various precipitates which act as hindrance to the dislocation movement. That is why, full recrystallization only occurs in the highly strained regions. It is not possible in the normal rolling condition as the energy required for this whole process is not present in the normal solution treatment condition to overcome the dispersoid pinning effect [35]. Similar behavior was observed for other aluminum alloys [35–37].

XRD analysis

Phases present in the homogenized sample have been identified using XRD analysis. XRD pattern of the homogenized sample confirms the presence of Al, MgZn2, Al3CuMg and Al3Mg2Zn1 phases labeled in figure 5. Other phases obtained from the EDS analysis such as MgSi and AlFe do not show any peak in the diffraction pattern. Their small amount may have restricted their detection in the XRD experiment.

DSC analysis

In this experiment, DSC analysis of this studied alloy at various conditions such as cast, homogenized and rolled is conducted. All three DSC curves are superimposed in a single plot, shown in figure 6. For all the conditions, a major peak is seen at around 451 °C. According to literature, a high endothermic peak for the dissolution of η(MgZn2) phase occurs between 300 °C and 490 °C [38]. XRD and EDS analysis already confirmed the presence of η phase in the alloy. So, this major peak is due to the dissolution of η phase. Dissolution of this η phase is completed around 460 °C. If the cast product is quickly heated above 450 °C, constituents will melt causing over burning. That is why the homogenization temperature of this type of alloy should be below 450 °C to avoid incipient melting or over burning [39, 40]. Another endothermic peak around 500 °C is seen in the cast and homogenized samples. These peaks are due to the dissolution of S (AlCuMg) phase [41]. It is seen that the peak intensity decreases in the rolled product. This weaker endothermic peak at around 500 °C indicates the decrease of the fraction of S phase in the rolled product.

Flow behavior

True stress strain curve of a material shows real condition of plastic deformation. The curve is highly dependent on the condition of strain rate and deformation temperature. In a given strain rate and temperature condition, initially the stress of material increases with the increase of strain in the elastic region due to rapid dislocation multiplication. With the increased straining, elastic deformation converts to plastic deformation due to more dislocation multiplication and energy accumulation. The accumulated energy after crossing the threshold value can provide energy for dislocation movement. This energy then initiates dynamic softening mechanisms such as dynamic recovery and dynamic recrystallization preventing further increase of flow stress. With further straining, dynamic softening mechanism prevails, flow stress decreases more and stress concentration increases. At some point, concentrated stress becomes so high that internal voids begin to coalesce and grow and finally fracture occurs. Thus the hot tensile deformation of a material is a combined action of work hardening, dynamic
recovery and dynamic recrystallization. With the change in the condition of strain rate or deformation temperature, basic mechanism of deformation remains same. They only galvanize the mechanism. At high strain rates, rapid dislocation multiplication occurs causing work hardening. Work hardening plays the major role in increasing peak stress. At low strain rate, a material get enough time for energy accumulation and higher temperature increases the energy for dislocation movement and dislocation annihilation. Both these conditions promote dynamic softening and reduce peak stress by recovery and recrystallization process.

A series of true stress–strain curves have been found from the tensile tests conducted at various temperatures and strain rates (figure 7). From these true stress–strain curves, it is seen that, the peak stress or flow stress increases with the increasing strain rate or decreasing deformation temperature.

Maximum elongation before fracture is an important property to determine the deformability of a material. Variation of maximum % elongation is shown with respect to the changes in the deformation temperature and strain rate in figure 8.

From figure 8, it is apparent that, increase of deformation temperature increases elongation to fracture. Under normal straining condition, an edge dislocation moves only in the slip plane that contains the dislocation line and its burger vector by gliding. However, increased temperature causes the edge dislocation to climb directly above or below the slip plane to be rearranged and produce low angle grain boundaries [42]. Diffusion of vacancies and interstials only play the major role in this dislocation climbing. So, it can be said that dislocation climb is a diffusion controlled process. Since the increase of temperature increases the rate of diffusion of any system, diffusion climb occurs more readily in the elevated temperature [31]. New slip systems also become active when a metal is deformed at high temperature [31]. Again, grains of polycrystalline material can also move by grain boundary sliding at elevated temperature by a shear process [43]. Elevated temperature thus promotes increased deformation of this studied alloy by slip, activation of multiple new slip systems, sub-grain formation and grain boundary sliding. They all made the material movement easier causing increased elongation before fracture. However, strain rate effect is not as simple as the temperature effect which goes well with the literature [44]. With the increase of strain rate from 0.000 05 to 0.000 35 s\(^{-1}\), elongation to fracture decreases. This is because, at higher strain rates, dislocations can get tangled making plastic deformation harder [1]. Lower strain rates also provide materials enough time for flowing and grain boundary sliding. When strain rate is further increased to 0.0005 s\(^{-1}\), elongation to fracture increases up to 350 °C and then decreases at 400 °C. One reason for this can be that when strain rate is relatively low, rapid agglomeration of voids or cracks can occur causing the material to fail catastrophically. But at higher temperature, temperature effect plays a strong role and makes the material to follow general rule [45] of strain rate and elongation. That is why; elongation to fracture at 0.000 35 s\(^{-1}\) and 400 °C is much greater than the elongation at 0.0005 s\(^{-1}\) strain rate and the same temperature. In this studied alloy, highest elongation to fracture was obtained at 0.000 05 s\(^{-1}\) strain rate and 400 °C temperature.

**Constitutive equations of flow stress**

An accurate model of flow stress can predict flow and peak stress at different deformation conditions. This is useful for selecting the working condition of a material such as what should be the highest temperature or strain rate that the material can withstand without failing catastrophically at a certain condition. Among all the models, Arrhenius model has been proved to be the most accurate one in predicting flow or peak stress. In this section, a constitutive model following Arrhenius-type equation will be used for the stress prediction of Al–9.6Zn–3.1Mg–1.84Cu alloy.

During hot deformation of materials, relations among temperature, strain rate and flow or peak stress of a material can be expressed as [15, 46–51]

\[ \dot{\varepsilon} = A_1 \sigma^n \exp \left( -\frac{Q}{RT} \right) \quad \text{[when } \alpha \sigma < 0.8] \ldots, \]  \hspace{1cm} (1)

\[ \dot{\varepsilon} = A_2 \exp(\beta \sigma) \exp \left( -\frac{Q}{RT} \right) \quad \text{[when } \alpha \sigma > 1.2] \ldots. \]  \hspace{1cm} (2)

Here \( A_1, A_2, n_1, \beta \) are all constants. These two equations require special stress condition. Sellars and Mctegart [52] have proposed a more general form of equation that can be used in a wide range of stress condition. This hyperbolic sine type equation is expressed below,

\[ \dot{\varepsilon} = A \sinh (\alpha \sigma)^n \exp \left( -\frac{Q}{RT} \right) \quad \text{[for all } \sigma] \ldots. \]  \hspace{1cm} (3)

Here, \( a = \frac{\beta}{n_1}. \) \( \beta \) is obtained from the average slopes of the curves \( \ln \dot{\varepsilon} \) versus \( \sigma \) and \( n_1 \) is obtained from the average slopes of the curves \( \ln \dot{\varepsilon} \) versus \( \ln \sigma. \)
Zener–Holomon parameter $Z$ again can be expressed from an exponential equation which is

$$Z = \dot{\varepsilon} \exp\left(\frac{Q}{RT}\right)$$

(4)
or,

\[ Z = A [\sinh (\alpha \sigma)]^n \text{[from eqn 3].} \ldots. \] (5)

Now, taking logarithm on the both sides of the equation (3),

\[ \ln \varepsilon = \ln A + n \ln[\sinh (\alpha \sigma)] - \frac{Q}{RT} \]

or,

\[ \ln [\sinh (\alpha \sigma)] = \frac{\ln \varepsilon}{n} + \frac{Q}{RT} - \frac{\ln A}{n} \ldots. \] (6)

The value of \( n \) can be obtained by differentiating equation (6) taking temperature constant. Thus the value of \( n \) can be obtained from the average slope of the \( \ln \varepsilon \) versus \( \ln[\sinh (\alpha \sigma)] \) curves

\[ n = \frac{\partial \ln \varepsilon}{\partial \ln[\sinh (\alpha \sigma)]} \bigg|_{\varepsilon}. \]

If strain rate is taken constant, then differentiating equation (5) gives the value of \( S \) which is the average slope of the \( \ln [\sinh (\alpha \sigma)] \) versus \((1/T)\) curves

\[ S = \frac{\partial \ln [\sinh (\alpha \sigma)]}{\partial \left(\frac{1}{T}\right)} \bigg|_{\varepsilon}. \]

So the value of \( Q \) can be expressed as

\[ Q = nRS \ldots \] (7)

In all temperature and strain rate condition, the values of peak stresses are shown in Table 4.

From the slopes of the curves of \( \ln \varepsilon \) versus \( \sigma \) and \( \ln \varepsilon \) versus \( \ln \sigma \) in figure 9, the values of \( \beta \) and \( n_1 \) is obtained.

From the slopes of \( \ln \varepsilon \) versus \( \sigma \) and \( \ln \varepsilon \) versus \( \ln \sigma \) from figure 9, the value of \( \beta \) is found to be 0.236 and \( n_1 \) is found to be 23.33. The value of \( \alpha \) is \( \frac{\beta}{n_1} \) which is 0.01 MPa\(^{-1}\). Taking the \( \alpha \) value, \( \ln [\sinh (\alpha \sigma)] \) versus \( \ln \varepsilon \) and \( \ln [\sinh (\alpha \sigma)] \) versus \( T^{-1} \) plots are drawn in figure 10.

| Strain rate \( s^{-1} \) | 250 °C | 300 °C | 350 °C | 400 °C |
|-------------------------|--------|--------|--------|--------|
| 0.00005                 | 160.62 | 105.2151 | 76.73883 | 25.07363 |
| 0.00015                 | 181.4691 | 130.5678 | 100.96 | 35.46457 |
| 0.0005                  | 220.21 | 162.3125 | 120.589 | 47.1242 |

Table 4. Peak stress values at various temperature and strain rate condition.
The value of material constant $n$ and $s$ can be obtained from the slopes of the curves in figure 10. Taking the average slopes from figures 10(a) and (b) and putting these value in equation (7), activation energy is found to be $146.2 \text{ kJ mol}^{-1}$ which matches with a similar kind of alloy [53].

Putting the value of $Q$ in equation (4) at all temperature and strain rate condition, $Z$ values are obtained. Taking logarithm on the both side of the equation (5),

$$\ln Z = \ln A + n \ln[\sinh (\alpha \sigma)].$$

Figure 4 shows the $\ln Z$ versus $\ln[\sinh (\alpha \sigma)]$ plots. The value of $A$ is found from slope of figure 11. The value of $A$ is found to be $1.23 \times 10^7$. The material constants for the studied alloy is shown in table 5. Similar kind of data has also been found in the literature [1].

The equations for finding Zener–Holomon parameter and the relation among strain rate, deformation temperature and peak stress for this Al–9.6Zn–3.1Mg–1.84Cu are established below:

$$Z = \varepsilon \exp\left(\frac{Q}{RT}\right),$$

$$\varepsilon = 5.98 \times 10^8[\sinh (0.01\sigma)^{0.64}] \exp\left(-\frac{146200}{RT}\right),$$

$$\sigma = \frac{1}{0.01} \sinh^{-1}\left(\frac{Z}{5.98 \times 10^8}\right)^{1.64}.$$

By following these equations, peak stress of this material can be predicted under different deformation temperature and strain rate. Accuracy of this constitutive equations are determined by comparing predicted peak stress ($\sigma_p$) with the peak stress obtained from the experiment ($\sigma_e$), the formula for determining relative
Figure 10. Relationships of (a) $\ln[\sinh(\alpha\sigma)]$ versus $\ln \varepsilon$ and (b) $\ln[\sinh(\alpha\sigma)]$ versus $T^{-1}$.

Figure 11. Relationships of $\ln Z$ versus $\ln[\sinh(\alpha\sigma)]$. 
error is,

\[
\text{Relative error} = \frac{\sigma_p - \sigma_e}{\sigma_e} \times 100\%. 
\]

Relative error value is shown in table 6. From the table, it is seen that, the equations have good prediction accuracy at various deformation conditions. From table 6, it is seen that the maximum relative error is 4.88%. The predicted and the experimental values are very close to each other. So it can be said that these equations are applicable to predict the flow stress of this high zinc containing hot rolled alloy samples.

**Microstructural evolution**

Optical micrographs of the studied alloy in the temperature range of 250 °C–400 °C at 0.00005 s\(^{-1}\) strain rate are shown in figure 12. At 250 °C, recrystallized grains begin to form at the grain boundary and at 300 °C, whole structure contains fine, equiaxed recrystallized grain with low dislocation densities. Hot deformation at sufficiently elevated temperature provides the energy required for atom and dislocation movement. As a result, merging of sub grains and conversion of low angle grain boundaries to high angle grain boundaries occurs by reducing the number of dislocations and causing subgrain growth. Grain boundary sliding also occurs at the existing grain boundaries. With continuous deformation, continuous dynamic recrystallization and sub grain growth occurs due to deformation induced continuous reaction. Eventually, this subgrain growth forces the high angle boundaries to migrate and grain growth occurs.

The driving energy for this process was the energy difference between strained and unstrained material. It is seen that, the whole structure is consumed at 300 °C to form recrystallized grain. With the further increase of temperature, grain growth occurs to reduce energy. When grain growth occurs, grain boundary decreases and so the energy decreases.

Generally, SFE is the most important thing that influences dynamic recovery and recrystallization. SFE determines the amount of dislocation slip and climb that would occur in a material during metal working. Materials with high SFE can have rapid dislocation climb and so, dislocation mobility increases and a significant amount of recovery can occur. However, low SFE materials face difficulty in dislocation climbing and cross slip and lower amount of dislocation can accumulate on a slip plane. So, they exhibit a small amount of recovery before recrystallization. Pure Al has high SFE of about 160–200 mJ m\(^{-2}\) [54]. But various alloying addition generally reduces SFE [55]. In the studied alloy, high amount of alloying element may have reduced SFE and so the recovery before recrystallization was not observed here. Again at such low strain rate, alloy gets more time for recrystallization rather than recovery.

Figure 13 shows hot deformed microstructures at 400 °C temperature and different strain rates. It is seen that, with the increase of strain rate, serrations begins to develop at the grain boundary. At higher strain rates, fine grains co-exist with the existing elongated grain boundaries and at 0.0005 s\(^{-1}\) strain rate, number of fine

### Table 5. Material constant values for Al–9.6%Zn–3.1% Mg–1.84%Cu alloy.

| \(A\) s\(^{-1}\) | \(\alpha\) MPa\(^{-1}\) | \(n\) | \(Q\) kJ mol\(^{-1}\) |
|----------------|-----------------|-----|-------------|
| 5.98 \times 10^8 | 0.01            | 3.64| 146.2       |

### Table 6. Comparison between predicted peak stress and experimental peak stress.

| Temperature °C | Strain rate s\(^{-1}\) | Predicted peak stress, \(\sigma_p\) | Experimental peak stress, \(\sigma_e\) | Absolute relative error % |
|----------------|------------------------|-----------------------------------|-------------------------------|--------------------------|
| 250            | 0.00005                | 154.69361                         | 160.62                        | -3.68969                 |
|                | 0.00015                | 173.83795                         | 181.469                       | -4.20516                 |
|                | 0.0005                 | 209.49004                         | 220.21                        | -4.86806                 |
| 300            | 0.00005                | 103.84129                         | 105.2151                      | -1.30572                 |
|                | 0.00015                | 127.12888                         | 130.5678                      | -2.63382                 |
|                | 0.0005                 | 156.24678                         | 162.3125                      | -3.73707                 |
| 350            | 0.00005                | 77.4772                           | 76.73883                      | 0.96219                  |
|                | 0.00015                | 99.92163                          | 100.96                        | -1.0285                  |
|                | 0.0005                 | 117.97222                         | 120.589                       | -2.17                    |
| 400            | 0.00005                | 26.13037                          | 25.07363                      | 4.2145                   |
|                | 0.00015                | 36.04108                          | 35.46457                      | 1.6261                   |
|                | 0.0005                 | 49.42656                          | 47.1242                       | 4.88573                  |
grains increases compared to the other two strain rates. Grains at lower strain rate are also bigger due to the sufficient time for the grain growth at the deformation condition.

Generally, the value of Zener–Hollomon parameter ($Z$) gives an idea about the flow softening mechanism. With the decrease of $Z$ value that is with the increase of temperature and decrease of strain rate, dynamic
recovery transforms into continuous dynamic recrystallization. Variation of lnZ value with temperature and strain rate is shown in table 7. It is seen that with increasing temperature and decreasing strain rate, Z value decreases and thus the softening mechanism changes accordingly.

Conclusion

In this experiment, hot tensile behavior of a hot rolled Al–9.6%Zn–3.1%Mg–1.84% Cu alloy has been investigated using uniaxial tensile testing under various conditions. The effect of different deformation temperature and strain rate condition on the nature of deformation has been elaborately explained and compared them with the deformed microstructures. Depending on the results from the true stress–strain curves, constitutive equations have been established to predict the peak stress of the studied material under different deforming conditions. Important findings of this investigation are summarized below:

• With the increase of temperature and decrease of strain rate, peak stress of this material decreases as it should generally occur. This behavior can be explained by the Zener–Hollomon parameter (Z) and hyperbolic sine equation. Activation energy has been found to be 146.2 kJ mol\(^{-1}\).

• With the decrease of Z value, deformation mechanism transforms to dynamic recovery from dynamic recrystallization because high temperature and low strain rate favors the recrystallization condition.

• Accuracy of this developed model is very good with lower that 5% relative error at various deformation conditions.

Acknowledgments

This study has been financially supported by a subproject CP 3117 under the higher education and quality enhancement project (HEQEP) of UGC, Bangladesh.

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Table 7. ln(Z) values of the deformed alloy.

| Temperature (°C) | Strain rate (s\(^{-1}\)) | 0.00005 | 0.00015 | 0.0005 |
|------------------|-------------------------|--------|--------|--------|
| 250              | 25.02                   | 26.58  | 27.79  |
| 300              | 21.98                   | 23.49  | 24.7   |
| 350              | 19.42                   | 20.9   | 22.11  |
| 400              | 17.24                   | 18.7   | 19.89  |
