Impact of high strain rate deformation on the mechanical behavior, fracture mechanisms and anisotropic response of 2060 Al-Cu-Li alloy

Ali Abd El-Atya,b, Yong Xu a,c,b, Shi-Hong Zhang a, Sangyul Ha d, Yan Ma a, Dayong Chen a

a Institute of Metal Research, Chinese Academy of Sciences, Shenyang 110016, PR China
b School of Engineering Science, University of Chinese Academy of Sciences, Beijing 100049, PR China
c School of Materials Science and Engineering, Nanjing University of Science and Technology, Nanjing 210094, PR China
d Corporate R & D Institute, Samsung Electro-Mechanics, Suwon 443-743, Republic of Korea

Highlights

- The mechanical behavior of AA2060 was investigated under HSR deformation and at room temperature.
- A novel gripping method was designed to prevent the distortion of strain waves during HSR experiments.
- The ductility of AA2060 was enhanced due to the adiabatic softening and inertia effect.
- The fracture behavior of AA2060-T8 was changed from brittle to ductile behavior under HSR deformation.
- Johnson-Cook constitutive model was modified to predict the dynamic flow behavior of AA2060-T8.

Abstract

Since AA2060-T8 was introduced in the past few years, investigating the mechanical response, fracture mechanisms, and anisotropic behaviour of AA2060-T8 sheets under high strain rate deformation has been crucial. Thus, uniaxial tensile tests were performed under quasi-static, intermediate, and high strain rate conditions using universal testing machines as well as split Hopkinson tensile bars. The experimental results showed that the ductility of AA2060-T8 sheets was improved during high strain rate deformation because of the adiabatic softening and the inertia effect which contribute to slow down the necking development, and these results were verified by the fracture morphologies of high strain rate tensile samples. Furthermore, the strain rate hardening influence of AA2060-T8 was significant. Therefore, the Johnson–Cook constitutive model was modified to consider the effects of both strain and strain rates on the strain hardening coefficient. The results obtained from the improved Johnson–Cook constitutive model are in remarkable accordance with those obtained from experimental work. Thus, the improved Johnson–Cook model can predict the flow behavior of AA2060-T8 sheets at room temperature over a wide range of strain rates. The results of the present study can efficiently be used to develop a new constitutive model.
Introduction

Recently, the family of third generation Al-Li alloys has shown promise as materials for the components used in aerospace, aircraft, and military applications because of their remarkable mechanical and physical properties, such as low density, good corrosion resistance, and high specific strength and stiffness [1]. These outstanding mechanical and physical properties are mainly caused by adding Li to the Al matrix. For instance, adding 1 wt% Li increases the elastic modulus and reduces the density of alloys by approximately 6% and 3%, respectively [1,2]. In 2011, Alcoa Corporation launched AA2060-T8 as a new third generation Al-Li alloy to supersede AA7075-T6 and AA2024-T3 for fuselage and lower and upper wing structures [1]. Although the AA2060-T8 alloy demonstrates remarkable mechanical and physical properties, it displays poor formability at room temperature, which hinders its broad application [2].

Since AA2060-T8 was launched a few years ago, few investigations on studying the deformation behavior and determining the relationship between the mechanical response and the texture of this alloy have been performed. For example, Abd El-Aty et al. [3] studied the tensile properties of AA2060-T8, AA8090, and AA1420 sheets at room temperature and quasi-static strain rates. They found that the tensile properties of these alloys did not display a constant trend with increasing strain rate, and they recommended investigating the dynamic behavior of these alloys under high strain rates and various loading orientations. Thereafter, Abd El-Aty et al. [4] proposed a novel methodology called ‘computational homogenization-based crystal plasticity modelling’ and established a multi-scale constitutive model to link the mechanical response of AA2060 with the microstructural states. These authors used this novel methodology and the proposed constitutive model to predict the mechanical response and texture evolution and to capture the anisotropic responses of AA2060-T8 at room temperature and quasi-static strain rates [4–6]. Ou et al. [7] studied the hot deformation behavior of AA2060 and reported that the main reason for softening during hot forming is dynamic recovery. In addition, they found that the optimum hot working conditions lie within the strain rate and temperature ranges of 0.01–3 s⁻¹ and 380–500 °C, respectively. Gao et al. [8] investigated the practicability of manufacturing aircraft components from AA2060 using hot forming and in-die quenching (HFQ) process. They found that the optimum temperature and strain rate to manufacture these parts from AA2060 are 470 °C and 2 s⁻¹, respectively. Jin et al. [9] proposed a pixel rotation method (PRM) to investigate the texture evolution and mechanical behavior of AA2060-T8 during bending process. They characterized the texture contents in the bent specimens with different radii (using PRM) and noticed that the mechanical strength of AA2060-T8 was improved in the longitudinal direction (i.e. the specimen axis parallel to the rolling direction); 45° to the rolling direction, and long–transverse direction (i.e. the specimen axis perpendicular to the rolling direction) with reduced bending radius. These improvements in the mechanical strength in these three directions are attributed to the strain hardening during bending, since, a large number of dislocations are generated and accumulated during plastic deformation and this increase in dislocation density lead to work hardening during bending. The amount of low-angle grain boundaries (LAGBs) is the main manifestation of dislocation density. LAGBs are a crucial parameter for the characterization of deformation degree. The greater the applied deformation, the higher the proportion of LAGBs will be. The proportion of LAGBs increased due to the subdivision of grains during bending. The amounts of LAGBs obviously increased with decreasing bending radii. Thus, mechanical strength was increased with reduced bending radius. Subsequently, Jin et al. [10] investigated the dislocation boundary structures of AA2060 during the bending process and found that three types of microstructures were formed. Later, Jin et al. [11] analysed the damage mechanisms and the microstructure evolution of AA2060-T8 during bending using in-situ bending test. They loaded the test-samples (bending samples) with a series of punches of different radii and used digital image correlation and electron backscatter diffraction techniques as well as scanning electron microscopy for microstructure and texture evolution. Their results showed that the strain localization in the outer surface (free surface) of the bending samples actuated damage to the microstructure. At the beginning of bending, crack initiation occurred on the free surface with maximum strain, and the shear crack propagated along the macro-shear band.

According to the above discussion, the dynamic deformation behavior of AA2060-T8 under high strain rate conditions has not yet been investigated. High deformation rate or high speed forming is considered as a significant method to improve the formability of lightweight metallic materials which have poor formability at room temperature [12]. This phenomenon is very interesting and important in sheet metal forming, thus, it is valuable to explore the mechanical response, fracture mechanism, and flow behavior of AA2060-T8 sheets under high deformation rate. Furthermore, investigating the anisotropic coefficient under high deformation rate is also meaningful to quantify the thinning resistance of the AA2060-T8 sheets under high strain rate deformation. These investigations can efficiently be used to develop a new manufacturing route based on impact hydroforming technology (IHF) to manufacture sound thin-walled-complex shape components from AA2060-T8 sheets at room temperature.

The flow behaviors of Al and Al-Li alloys under high speed conditions are complicated because they depend on several factors, such as the deformation mode, strain, and strain rates [13,14]. These factors control strain hardening, which in turn affects the flow behavior and formability of Al and Al-Li alloys [13]. Therefore, predicting the flow behavior of AA2060 sheets under a wide range of strain rates is crucial. Constitutive equations are usually used to predict the flow behavior of materials in a form that can be used in finite element (FE) codes to simulate the mechanical response of materials under different forming conditions [13–17]. These constitutive models include physically based constitutive models, phenomenological constitutive models, and artificial neural network (ANN)-based modelling [13]. Basically, the optimal constitutive model should possess a moderate number of material parameters, which can be assessed via a small amount of experimental data, and be able to accurately predict the mechanical behavior of materials over a wide range of rheological variables [13,14]. Physically based models may afford exact representation of the flow behavior of materials over a wide range of rheological variables [17]. Furthermore, they can trace the microstructural evolution by using the dislocation density as a variable, in which the constitutive equations based on dislocation theory may correctly characterize the effects of strain hardening and dynamic softening [18–23].
Nevertheless, physically based models are not usually preferred because they require a large amount of data from accurate experiments and a large number of material properties and constants that might not be available in the literature [13,24]. Phenomenological constitutive models do not require a full understanding of the rheological variables included in the forming process, in which the constitutive equation can be determined by fitting and regression analysis [23,24]. Hence, these models are widely used to predict the flow behavior of materials over a wide range of temperatures and strain rates [21–27]. Furthermore, they can be integrated into FE codes to simulate actual forming processes under different forming conditions. However, they cannot link the microstructural state of materials with their mechanical behavior, which is not crucial in the current investigation [13,17].

Accordingly, the objectives of this study are to investigate the mechanical response, fracture mechanisms, and anisotropic behavior of AA2060-T8 sheets under high strain rate deformation. Additionally, a phenomenological constitutive model has been developed to predict the flow behavior of this alloy under quasi-static (QSR), intermediate (ISR), and high (HSR) strain rate conditions; thus far, no applicable constitutive model to predict the mechanical behavior of AA2060-T8 under a wide range of strain rates has been proposed.

Experimental material and procedures

Material description

The material used in this study was rolled sheets Al-Cu-Li alloy 2060-T8 sheet (T8: solution heat treated, then cold worked and finally, artificially aged). The chemical composition, and the microstructure of as-received AA2060-T8 sheet are presented in Table 1 and Fig. 1a, respectively. The samples used for microstructure characterization were cut in rolling direction (RD), ground by silicon Carbides (SiC) papers, polished through diamond pastes, and etched via the solution of Keller’s reagent (85% H2O, 3% HF, 6% HNO3, and 6% HCl). As depicted in Fig. 1a, it was observed that the grains exhibited a typical pancake-shaped grain structure which is the evident that AA2060-T8 sheets display a typical cold-rolled microstructure. Furthermore, the grains are significantly elongated and flattened in RD, and the grain sizes are relatively large compared with other Al and Al-Li alloys. Most of Al and Al-Li alloys manifest the initially anisotropic textures due to the thermomechanical processes in which the deformation history is generally unknown [1]. Thus, in this study, HKL Channel 5 Electron backscatter diffraction (EBSD) analysis system was used to characterize the grain size and texture of the AA2060-T8 samples. The samples used for EBSD for characterization (with upper surfaces of RD × ND) were first mechanically ground by SiC papers, thereafter, electro-polished in HClO4:CH3OH (10:90, by volume) solution at room temperature under an applied voltage of 20 V for 15–20 s. The texture components, such as Goss, Brass, Cube, Copper, and S were detected within 15° of the nearest ideal component. For simulation reason, the initial crystallographic data obtained from the EBSD measurement was reduced by the coarsening technique that removes the pixel every two pixels and reduces the number of points in a dataset by a factor of four. This method was repeated to obtain 50 crystallographic orientations which approximate the initial texture of the AA2060-T8 specimen. The (1 1 1) pole figure of the reduced texture of as-received AA2060 sheet is depicted in Fig. 1b.

Uniaxial tensile experiments

Thus far, perfectly describing the mechanical behavior under a wide range of strain rates using one testing machine is impractical because of the restricted range of the velocity of these machines. Thus, tensile experiments are divided into quasi-static, static, and dynamic experiments based on the magnitudes of the strain rates [28], as depicted in Fig. 2 and summarized in Table 2 [28–30]. Therefore, in the current investigation, three different uniaxial tensile experiments were performed to describe the mechanical behavior of AA2060-T8 sheets at QSR, ISR, and QSR, as listed in Table 3. The tensile samples used at HSR, ISR, and QSR were all sheets (t = 2 mm).

Uniaxial tensile tests at QSR and ISR

A 100 kN Instron 5980 and a 150 kN Zwick/Roell proline Z150 were used to carry out the tensile tests at room temperature and at QSR (0.001–0.1 s−1) and ISR (1 s−1), respectively, as presented in Table 3. The setup of both the QSR and ISR experiments and the dimensions of the specimens used in these experiments are shown in Fig. 3a and b, respectively. To study the mechanical response and flow behavior of the AA2060-T8 sheet at QSR and ISR, the tensile specimens were cut using an electrical discharge machine in the RD of the sheet. Furthermore, to investigate the anisotropic behavior of AA2060-T8, the specimens were machined in five directions at 0°, 30°, 45°, 60°, and 90° (transverse direction) with respect to the RD, as depicted in Fig. 3c. Each test condition was studied at least three times to ensure consistency and repeatability. The average values of these three repetitions were considered; thus, every experiment affects the constitutive fitting. Furthermore, each experiment contains an equal amount of data and is hence weighted equally.

Uniaxial tensile test at HSR

HSR tensile tests were performed using the split Hopkinson tensile bars (SHTB) apparatus to investigate the dynamic behavior of the AA2060-T8 sheet in the RD at room temperature and different strain rates as listed in Table 3. The effect of the sample orientation in the HSR tensile tests was not considered since the sample orientation has a significant impact in the case of QSR and ISR but not HSR [1,3,11]. Furthermore, the results obtained from both QSR and ISR tensile tests were enough to investigate the influence of sample orientation on the tensile properties of AA2060-T8 sheets and characterize the degree of in-plane anisotropy. However, investigating the influence of HSR deformation on the anisotropic coefficient (r-value) is crucial to quantify thinning resistance of AA2060-T8 sheets.

The SHTB apparatus used in this investigation was consisted of three bars named the projectile or striker (with a maximum velocity of 80 m/s), incident bar (input bar), and transmitted bar (output bar), as well as strain gauges, amplifiers, and an oscilloscope as depicted in Fig. 4a and b. These three bars are free to slide and
supported by adjustable holders to ensure good alignment. Furthermore, the cross section areas of the striker and incident bars were designed to be identical to avoid impedance mismatch between them. The most critical issue of HSR tensile tests using SHTB apparatus is controlling and increasing the strain rate. This leads some researchers [31–37] to develop the setup of the SHTB apparatus to increase and control the strain rate, meanwhile keep the test design simple and have the possibility to directly compare the results with those acquire at lower strain rates. Generally, very high strain rates can be obtained using (SHTB) apparatus by two ways. The first way is to increase the speed of the striker bar, however, this leads to increase the stress level in the striker bar, which is restricted by the yield strength of the sticker bar’s material. Thus, the second way was used in the current study. The second way depends mainly on controlling and reducing the dimensions of the tensile sample, because to-date, the samples used for HSR tensile testing by SHTB apparatus did not have a standard design and geometry. Thus, designing HSR tensile sample is a significant aspect of the current study. Nevertheless, there are some aspects should be considered when designing the HSR tensile sample. For instance, the gauge length of the sample should be small to reduce the ring-up time and inertial effects, meantime, the sample should be large enough to be representative of the material behavior under HSR testing. Furthermore, the ratio between the gauge

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**Table 2**

| Strain rate               | Magnitude               |
|---------------------------|-------------------------|
| Quasi-Static (QS) and low strain rates | $10^{-3} \leq \varepsilon < 10^{-1}$ |
| Intermediate or medium strain rates | $10^{-1} \leq \varepsilon < 10^{2}$ |
| High strain rate (HSR)    | $10^2 \leq \varepsilon \leq 10^4$ |
| Ultra HSR                 | $10^4 < \varepsilon$    |
length and width of the tensile sample (length/width) must be considered when reducing the gauge length of the tensile sample to ensure a uniaxial state of stress. Furthermore, the length/width ratio of the HSR sample must be almost the same to the QSR and ISR samples to ascertain that the results obtained from HSR sample could be compared with that obtained from QSR and ISR samples without particular size effects on the material response [1,31–37].

Based on the aforementioned aspects, a new HSR tensile sample was designed to achieve very high strain rates and perform HSR tensile test correctly. This design followed the mechanical response of a standard ASTM specimen, while meeting the requirements for specimens used in dynamic experiments.

The HSR experiment was supposed to be started once the tensile sample was placed between the incident and transmitted bars. However, the material being studied was rolled sheets with a thickness of 2 mm. Thus, a novel gripping method (clamp) was also designed to integrate the HSR tensile sample into the SHTB apparatus to provide adequate clamping forces to avoid tensile specimens from slipping during the experiments and to introduce a low mechanical impedance to prevent distortion of the waves. The shape of the HSR tensile sample is depending on the design of the clamp. Therefore, many trials were performed to obtain the optimum shape and design of the tensile sample and clamp. The initial design of the novel clamp successfully avoided the slipping of the tensile specimen. Nevertheless, the waves were distorted, as shown in Fig. 4c. Thereafter, further modifications were made to the initial design of the tensile sample and clamp to prevent the tensile sample from slipping during the test as well as avoid distortion of the waves. Nonetheless, the waves were still distorted, as depicted in Fig. 4d and e. After that, additional modifications were carried out until adequate clamping forces were provided and the distortion of the waves was minimized, as depicted in Fig. 4f.

Once the novel clamp was implemented in the SHTB apparatus, the tensile specimen was placed between the incident and transmitted bars; thereafter, the striker situated on the incident bar impacted the flange, leading to the generation of a tensile wave (incident wave) that propagated along the incident bar, as depicted in Fig. 5a. The strain gauge located on the incident bar recorded the incident wave once it passed. The amplitude ($\sigma_i$) and length ($L_i$) of the incident wave were calculated as follows:

$$\sigma_i = \frac{1}{2} \times \rho_{\text{input}} \times E_{\text{input}} \times v_{\text{impact}}$$

$$C_{\text{input}} = \frac{E_{\text{input}}}{\rho_{\text{input}}}$$

$$L_i = 2L_{\text{striker}}$$

where $\rho_{\text{input}}$ and $E_{\text{input}}$ are the density and elastic modulus of the material of the incident bar, $C_{\text{input}}$ is the velocity of the longitudinal elastic wave of the incident bar, $v_{\text{impact}}$ is the impact velocity, and $L_{\text{striker}}$ is the length of the striker.

Once the incident wave hits the sample, it is partly reflected back ($\varepsilon_0$) through the incident bar and partly transmitted ($\varepsilon_T$) through the tensile sample and the transmitted bar, as shown in Fig. 5a. These reflected and transmitted waves were recorded by the strain gauges (using a high velocity acquisition system, i.e., an oscilloscope) situated on the incident and transmitted bars, respectively. A schematic and a real set of waves detected during the SHTB experiment are depicted in Fig. 5b and c.

By introducing the relationship between the particle velocity and the elastic strain waves, the displacements of both ends of the tensile specimen ($u_{\text{input}}$, $u_{\text{output}}$) were defined by

$$u = C \int_0^t \varepsilon(t) dt$$

Thus,

$$u_{\text{input}} = C_{\text{input}} \int_0^t \varepsilon_i(t) - E_{\text{input}} \int_0^t \varepsilon_e(t) = C_{\text{input}} \int_0^t e_i(t) - e_e(t) dt$$

$$u_{\text{output}} = C_{\text{output}} \int_0^t e_T(t) dt$$

where $e_i$ is the incident strain wave and $e_T$ is the reflected strain wave.

By similarity, the transmitted strain wave on the other side of the tensile specimen was given by

$$u_{\text{output}} = C_{\text{output}} \int_0^t e_T(t) dt$$

where $C_{\text{output}}$ is the velocity of the longitudinal elastic wave of the transmitted bar and $e_T$ is the transmitted strain wave. It is assumed that $C_{\text{input}} = C_{\text{output}} = C$ by considering that the incident and transmitted bars have the same material properties.

Thus, the instant strain ($\varepsilon$) in the specimen was calculated as follows:

$$\varepsilon(t) = \frac{u_{\text{input}}(t) - u_{\text{output}}(t)}{L_0} = \frac{C}{L_0} \int_0^t [e_i(t) - e_e(t) - e_T(t)] dt$$

where $L_0$ is the initial length of the tensile specimen.

At the equilibrium condition, the forces at the input (incident bar) and output (transmitted bar) sides are equivalent,

$$F_{\text{input}}(t) = F_{\text{output}}(t)$$

Using Hooke's law, $E = \sigma / \varepsilon$ and $\sigma = F / A$, Eq. (9) is expressed as

$$E_{\text{input}} \times A_{\text{input}} \times e_{\text{input}}(t) = E_{\text{output}} \times A_{\text{output}} \times e_{\text{output}}(t)$$

$$e_{\text{input}}(t) = e_i(t) + e_e(t)$$

$$e_{\text{output}}(t) = e_T(t)$$

Thus,

$$E_{\text{input}} \times A_{\text{input}} \times [e_i(t) + e_e(t)] = E_{\text{output}} \times A_{\text{output}} \times e_T(t)$$

where $E_{\text{output}}$ is the elastic modulus of the material of the transmitted bar, and $A_{\text{input}}$ and $A_{\text{output}}$ are the cross section areas of the incident and transmitted bars, respectively.

### Table 3

Uniaxial tensile experiments matrix, ($\checkmark$) implies that the test was done at these conditions.

| Strain rate ($s^{-1}$) | Category | 0° | 30° | 45° | 60° | 90° |
|------------------------|----------|----|----|----|----|----|
| 10–3                   | QSR      | $\checkmark$ | $\checkmark$ | $\checkmark$ | $\checkmark$ | $\checkmark$ |
| 10–2                   | $\checkmark$ | $\checkmark$ | $\checkmark$ | $\checkmark$ | $\checkmark$ | $\checkmark$ |
| 10–1                   | $\checkmark$ | $\checkmark$ | $\checkmark$ | $\checkmark$ | $\checkmark$ | $\checkmark$ |
| 1                      | ISR      | $\checkmark$ | $\checkmark$ | $\checkmark$ | $\checkmark$ | $\checkmark$ |
| 1733, 3098, 3651, 3919 | HSR      | $\checkmark$ | $\checkmark$ | $\checkmark$ | $\checkmark$ | $\checkmark$ |
Fig. 3. Experimental setup of (a) Instron 5980, (b) Zwick/Roell proline Z150 m/c used for tensile testing at QSR and ISR, respectively, and (c) The specimens cut in various directions with respect to RD.
It is assumed that $C_{input} = C_{output} = C$ based on the aforementioned assumption that the incident and transmitted bars have the same material properties and cross section area.

Therefore, Eq. (13) can be expressed as follows:

$$
\varepsilon_i(t) + \varepsilon_R(t) = \varepsilon_f(t)
$$

(14)

Once equilibrium verification was accomplished, the instant axial stress ($\sigma$) of the tensile specimen was calculated as follows:

$$
\sigma(t) = \frac{F_{input}(t) + F_{output}(t)}{2A_0}
$$

(15)

$$
\sigma(t) = \frac{|E_{input} \times A_{input} \times \varepsilon_i(t) + \varepsilon_R(t)| + |E_{output} \times A_{output} \times \varepsilon_f(t)|}{2A_0}
$$

(16)

Based on the aforementioned assumption that the cross section areas and the material properties of the incident and transmitted bars are similar $[A_{input} = A_{output} = A]$ ($E_{input} = E_{output} = E$), Eq. (16) was reduced to

$$
\sigma(t) = \frac{E \times A \times [\varepsilon_i(t) + \varepsilon_R(t) + \varepsilon_f(t)]}{2A_0}
$$

(17)

where $A_0$ is the cross section area of the tensile specimen.

For simplicity, the equilibrium condition was assumed to be valid during all the tests; thus, Eqs. (8) and (17), which are used to calculate the mean strain and mean stress, are generalized as follows:

$$
\varepsilon(t) = -\frac{2C}{L_0} \int_0^t \varepsilon_R(t) dt
$$

(18)

$$
\sigma(t) = \frac{E \times A}{A_0} \varepsilon_f(t)
$$

(19)

The instant axial strain rate ($\dot{\varepsilon}$) in the tensile sample was calculated from the first derivative of Eq. (18); thus, it can written as

$$
\dot{\varepsilon}(t) = \frac{v_{input}(t) - v_{output}(t)}{L_0} = -\frac{2C}{L_0} \varepsilon_R(t)
$$

(20)

Fig. 4. (a) The Schematic description, (b) The experimental setup of SHTB apparatus; (c) 1st, (d) 2nd, (e) 3rd, and (f) Final version of the novel clamp used to avert the specimens from slipping during the test.
Indeed, it was supposed to perform the HSR tensile tests at strain rate of 1500, 2500, 3500, and 4000 s\(^{-1}\) to investigate the dynamic behavior of AA2060-T8 sheets at most of HSR range (i.e. beginning, middle and end of HSR range). However, during the HSR tests, the strain rates are controlled by the speed of a striker bar and it is little bit difficult to control the speed of sticker bar. Thus, the speeds of the striker bar which equivalent to these range of strain rates are ranging from 10 to 35 m/s. The range of speed is based on the combination of the minimum speed of the SHPB set and the maximum impact velocity materials may reach in use. Furthermore, Eq. (20) indicates that with the SHTB apparatus, the tests are not performed exactly at a constant strain rate. Only in the ideal case of a perfectly rectangular reflected wave, i.e. a perfectly plastic response of the specimen, the strain rate is constant during the entire specimen deformation. In practice, this phenomenon is almost impossible to observe, and generally, the nominal strain rate (average value of the effective strain rate) is used to indicate the strain rate of tests performed on the SHTB apparatus.

Accordingly, the HSR experiments were performed at strain rates of 1733, 3098, 3651, and 3919 s\(^{-1}\). Each test condition was studied at least three times to ensure consistency and repeatability.

**Experimental results and discussion**

**Mechanical behavior and fracture morphologies under QSR and ISR**

The Engineering stress-strain (\(\sigma_e - \varepsilon_e\)) curves of AA2060-T8 under various loading directions at strain rates of 0.001, 0.01, 0.1 and 1 s\(^{-1}\) are depicted in Fig. 6a-d. These engineering stress-strains curves were obtained based on the initial cross sectional area of the tensile sample which changed during the test. Therefore, Eqs. (21) and (22) were used to convert the engineering stress-strain (\(\sigma_e - \varepsilon_e\)) curves of RD samples to true (\(\sigma_t - \varepsilon_t\)) curves to for precise constitutive fitting as shown in Fig. 7. These true (\(\sigma_t - \varepsilon_t\)) curves were plotted only up to ultimate tensile strength.
(UTS) points because beyond these points the diffuse necking occurs and the strain is not uniform in the tensile sample. Furthermore, the stress state deviates from uniaxial tension and shifts towards the plane-strain state once the UTS point is reached; thus, Eqs. (21) and (22) are no longer valid. The true \( \frac{\sigma_t}{C_0} \) curves of RD tensile samples under QSR and ISR deformation can be divided into elastic, yield, and hardening stages. The first stage is the elastic stage, where a linear relationship exists between the stress and strain. The Young’s modulus of the AA2060-T8 sheet obtained from the test results was 75 GPa. The second stage is the yield stage, where the strain rate has an obvious effect on the yield strength (YS), in which by increasing the strain rate from 0.001 to 1 s\(^{-1}\), the YS was increased as depicted in Fig. 7. The third and last stage is the hardening stage, where the AA2060-T8 sheet exhibits work hardening behavior, and the work hardening rate of these curves change with respect to strain rate.

\[
\sigma_T = \sigma_e (1 + \varepsilon_e) \tag{21}
\]

\[
\varepsilon_T = \ln(1 + \varepsilon_e) \tag{22}
\]

As shown in Fig. 6a–d and summarized in Fig. 8a and b, the UTS was also increased by increasing the strain rate; meanwhile, the elongation to fracture (EL\(_f\)) was decreased notably for the samples tested in the RD and at 90° w.r.t. the RD, which implies that the sample orientation has a significant impact on the mechanical behavior of AA2060-T8 sheets. Thus, the effect of sample orientation on the mechanical behavior of AA2060-T8 was investigated in this study. As depicted in Fig. 8a and b, under the same working conditions (room temperature and strain rate), the change in sample orientation from 0° to 60° w.r.t. the RD resulted in decreased YS and UTS, with a sharp increase in EL\(_f\), particularly for the samples at 45–60° with respect to the RD. For the sample orientations beyond 60°, the YS and UTS were increased, while EL\(_f\) was decreased. Thus, the tensile properties of AA2060-T8 sheets vary with respect to the loading direction, which signifies that the tensile properties of the AA2060-T8 sheet exhibit a serious degree of in-plane anisotropy. The differences in the YS, UTS and EL\(_f\) in AA2060-T8 sheets were attributed by the many factors such as the synergistic and independent interactive influences of the changes in the degree and nature of the crystallographic texture,
the recrystallization degree and the type and history of the deformation process before artificial ageing; and the fracture modes [1,2]. Thus, in the current study, the fracture morphologies of the tested samples were observed to investigate the fracture modes under different loading directions and strain rates. The anisotropy in tensile properties of AA2060-T8 sheets is very complex and not easy to straightforward analysis because it can be affected by many testing conditions (i.e. strain rates and working temperature) and parameters such as the strengthening phases, the precipitates in the microstructure, as well as the orientation, the sizes (widths and aspect ratios) and the shapes of grains and sub-grains. Thus, further investigation should be performed to link the anisotropic behavior of this alloy with the microstructural state.

Since the fracture morphology is reflective of the ductility and strength of the tensile samples, SEM was used to determine the fracture modes and describe the microscopic fracture features of the tensile samples tested under various loading directions and strain rates. The fracture modes of Al-Li alloys depend on various microstructural features, which are controlled by the alloying process, composition, and heat treatment procedures [3,6]. The common fracture modes and the features associated with them in Al-Li alloys are brittle intergranular fracture, cleavage of large constituent particles, “ductile” intergranular fracture, localized
transgranular shear fracture and dimpled transgranular fracture [3,37]. The stresses needed for these fracture modes vary from small for brittle intergranular fracture and cleavage of constituent particles to very large for transgranular fracture with deep dimples [37].

The typical fracture morphologies of the tensile samples tested in different loading directions, as well as under QSR and ISR, are shown in Fig. 9a–t. As depicted in these figures, the fracture morphologies of AA2060-T8 sheets displayed various characteristics that varied with the loading direction and strain rate. The fracture morphologies of the samples tested under loading directions of 45° and 60° with respect to the RD and at a strain rate of 0.001 s⁻¹ exhibited ductile fracture characterized by sizable dimples, as depicted in Fig. 9a and d. The volume fractions of the dimples in the fracture surfaces of the aforementioned samples (i.e., 45° and 60° w.r.t. the RD) were notably high compared to the fracture surfaces of the other samples tested at a strain rate of 0.001 s⁻¹ and under a loading direction of 30° with respect to the RD (which showed mixed ductile and brittle fracture, as shown in Fig. 9b) and samples tested under loading directions of 0° and 90° with respect to the RD (which showed the brittle fracture mode, as depicted in Fig. 9a and e, respectively). On the other hand, with increases in the strain rate to 0.01, 0.1, and 1 s⁻¹, the fracture modes appeared to be mixed ductile and brittle fracture for the loading directions of 45° and 60° with respect to the RD, as depicted in Fig. 9h, i, m, n, r, as well as brittle fracture for the samples tested at loading directions of 0°, 30°, and 90° with respect to the RD, as shown in Fig. 9f, g, j–l, o–q, and t. The aforementioned discussion is in good agreement with the tensile properties determined for these samples and substantiates the superior values of elongation obtained for the samples tested under loading directions of 45° and 60° with respect to the RD, as shown in Fig. 8a and b. A fracture window that summarizes the fracture modes of the tested samples under various loading directions and strain rates of 0.001–1 s⁻¹ is presented in Fig. 9u.

Mechanical response and fracture behavior under HSR

The engineering and true stress-strain curves of AA2060-T8 sheets under HSR conditions are depicted in Fig. 10a and b respectively. As aforementioned in previous section, the true stress-strain curves were plotted only up to UTS points. The mechanical behavior of AA2060-T8 sheets under QSR, ISR, and HSR were fundamentally similar, showing work hardening behavior, as depicted in Figs. 7 and 10. The work hardening behavior observed in HSR testing is more prominent than that detected in QSR and ISR testing. Thus, the YS, flow stress, and UTS at HSR were higher than their counterparts at QSR and ISR, which is generally attributed to the increased strain rates. Notwithstanding, the strain-hardening rate observed in HSR testing was lower than that in QSR and ISR testing. This is attributed to the competition between the strain hardening and thermal softening as a result of adiabatic temperature rise with increasing the strain rate. Thus, adiabatic softening influence is significant in HSR deformation and leads to abnormal mechanical behavior [34–37] in particular, when the strain rate increased to 1733, 3098, 3651, and 3919 s⁻¹ (i.e. strain rate changed from QSR to HSR). In addition, the elongation to fracture of AA2060-T8 sheets in the HSR zone was simultaneously improved by increasing the strain rate, which is an appealing feature in HSR deformation. This enhancement is attributed to the adiabatic softening with increasing the strain rate and the inertia effect which may contribute to diffuse necking, slow down the necking development and delay the onset of fracture. Therefore, high speed forming processes or HSR deformation processes can be considered as a significant forming technology for the manufacture of complex-shaped thin-walled components from Al-Li alloys at room temperature.

Since the fracture behavior can reflect both the ductility and strength of the tensile samples, we believe that the fracture behavior of AA2060 at HSR is certainly different from that at QSR and ISR. The fracture morphologies of the HSR tensile samples were characterized using SEM to investigate the fracture mode under HSR conditions. The typical fracture morphologies of AA2060-T8 HSR tensile samples are depicted in Fig. 11. As depicted in these figures, the fracture surfaces and morphologies of all AA2060-T8 HSR tensile samples contained various dimples and exhibited features typical of the ductile fracture mode. The fracture morphologies depicted in Fig. 9 and shows that the fracture surfaces of AA2060-T8 tensile samples tested at QSR and ISR were covered with very large and small shallow voids. The matrix among the neighbouring voids was approximately ruptured by the formation of small dimples. Thus, the diameters of the deep voids on the fracture surfaces decreased with increasing strain rate, which led to the brittle fracture mode, notably at high QSR (0.01 and 0.1 s⁻¹) as well as the beginning of ISR, as mentioned in the above section.

Fig. 8. (a) Yield (YS) and ultimate tensile stresses (UTS), and (b) Elongation to fracture (EL) of AA2060-T8 sheets at various strain rates (i.e. 0.001, 0.01, 0.1, and 1 s⁻¹) and sample orientations (i.e. 0°, 30°, 45°, 60°, and 90° with respect to the RD).
In contrast as shown in Fig. 11, in the HSR deformation area, we noted that the average dimple size of the HSR tensile samples increased with increasing strain rate, which implies that AA2060-T8 offers higher ductility and exhibits the ductile fracture mode under HSR deformation. In addition, the humps at the peak flow stresses of the mechanical behavior of AA2060-T8 sheets under HSR deformation became wider, which means that the alloy undergoes ductile fracture under HSR deformation. These results are in good accordance with the results observed in Fig. 10. The comparison between the fracture behavior of the tested tensile samples with reference to the strain rates is presented in the fracture window depicted in Fig. 9e. As shown in this figure, at high QSR (0.01 and 0.1 s⁻¹), AA2060 shows a brittle fracture behavior, and with increasing strain rate to high ISR and at the beginning of HSR, the behavior should transform to brittle – ductile behavior. Thereafter, with increasing strain rate in the HSR zone, AA2060 exhibits ductile fracture, as depicted in the fracture window in Fig. 11e.

The ductile fracture mode is divided into 3 stages, as depicted in Fig. 11f. The first stage is void nucleation, the second is void
growth, and the third is void coalescence. Initially, voids occur through the decohesion of the interface between the particles and the matrix; via the rupture of the particles during plastic deformation, the voids can grow until they connect together or coalesce to form continuous fracture paths, thus causing the final rupture of the specimens.

Anisotropic behavior of AA2060-T8 sheets under QSR, ISR, and HSR

As depicted in Fig. 6e and f, the tensile properties of AA2060-T8 sheets vary with respect to the loading direction, which means that the tensile properties of the AA2060-T8 sheets exhibit a serious degree of in-plane anisotropy. The degree of anisotropy in UTS is lower than the anisotropy in YS. The observed differences in YS and UTS were caused by synergistic and independent interactive influences of the changes in the degree and nature of the crystallographic texture; nature and distribution of strengthening phases; resultant microscopic deformation behavior, as well as, final heat-treatment condition and the degree of recrystallization [1]. The difference in the elongation to fracture was attributed to shearing of the Al₃Li precipitates and the resultant flow localization orientation with respect to the current stress states; the distribution and density of the intermediate-sized and coarse grains of the intermetallic particles; the type, distribution and morphology of the main strengthening phases, which are governing by alloying additions and thermo-mechanical processing; the recrystallization degree and the type and history of the deformation process before artificial ageing; the strength of grain boundaries; the width of precipitate-free zones; strength of grain boundaries, equilibrium phases densities along the grain boundaries and the fracture modes [1,2].

To quantify thinning resistance and influence of plastic anisotropy on the deformation and fracture behaviors of AA2060-T8 sheets, the anisotropic coefficient or Lankford parameter (r-value) was calculated. r-value is defined as the ratio of the true strain to the true thickness strain, as shown in Eqs. (23)–(26). On the one hand, for the QSR and ISR tensile tests, two independent extensometers were placed on the tensile specimens to simultaneously measure both the longitudinal (t_u) and width (t_w) strains. On the other hand, for the HSR tensile tests, the r-value was calculated using the method proposed elsewhere [3,31]. At HSR, t_u and t_w can be obtained from the grids printed on the surface of the tensile specimen. During the HSR test, due to the deformation, the shape of the rectangular grids continuously change, as depicted in Fig. 12a–d. The grid pattern is parallel to the direction of the uniaxial tension. A high speed camera was used to measure and detect gauge length deformations in the grid, and the plastic longitudinal, width, and thickness strains were calculated from the tested samples. The variations of the anisotropy (r-values) at QSR, ISR, and HSR are depicted in Fig. 12e.

\[
r = \frac{\varepsilon_w}{\varepsilon_t} \tag{23}
\]

\[
\varepsilon_t = -\left(\frac{\varepsilon_l + \varepsilon_w}{2}\right) \tag{24}
\]

\[
r = -\frac{\varepsilon_w}{\varepsilon_l} \tag{25}
\]

\[
r = \frac{\ln\left(\frac{\varepsilon_w}{\varepsilon_t}\right)}{\ln\left(\frac{\varepsilon_w}{\varepsilon_l}\right)} \tag{26}
\]

where \(\varepsilon_t\) is the thickness strain, \(x_1, x_2, y_1,\) and \(y_2\) are the lengths and widths of the rectangular grids along the X and Y axes before and after deformation.

If \(r < 1\), these materials easily thin and therefore have low biaxial strengths. In contrast, if \(r > 1\), the alloys have high thinning resistance because of the high strengths in the through-thickness direction and thus have high biaxial strengths [38]. Furthermore, deeper components with a smooth contour and minimum wrinkling can be achieved using a deep drawing process [39]. As shown in Fig. 12e, at QSR and at the beginning of ISR, the influence of the strain rate on the r-values seemed to be minimal; however, at ISR (strain rate of 1 s⁻¹), a slightly higher r-value was observed. At HSR, an apparent effect on the r-values was observed, in which increasing the strain rate in the HSR zone led to increased r-values. The higher r-values of AA2060-T8 sheets at QSR, ISR, and HSR make them more attractive not only for conventional metal forming processes but also for high speed forming processes, such as impact hydroforming, because of the ability of this alloy to resist thinning at HSR. Although the r-value can be used to evaluate the formability of sheets, studying the impact of the strain rate on the formability of sheet metals is difficult, particularly at a high rate of deformation. To determine the formability of AA2060-T8 sheets, performing formability tests at both quasi-static and high speeds, in addition to uniaxial tensile tests, is crucial.
Constitutive modelling

Original Johnson-Cook model

The QSR, ISR, and HSR deformation behavior of metallic materials can be characterized by different constitutive models that basically attempt to describe the dependency of the flow stresses on the strain, strain rate, and temperature, as presented in Eq. (27).

\[ \sigma = f(\varepsilon, \dot{\varepsilon}, T) \]  

(27)

These constitutive models include physically based constitutive models, phenomenological constitutive models, and ANN modelling [12]. Among these constitutive models (notably phenomenological models), the Johnson-Cook model (J-C) is often used for different metallic materials with different forming conditions and is available in most commercial FE codes [40,41]. For simplicity,
the materials are assumed to be isotropic to avoid the traditional concept of a yield surface in the constitutive equations [40]. Thus, the J-C model is presented as follows:

\[ \sigma = \frac{A + Be^n}{(1 + Cln\dot{\varepsilon}^\prime)} \begin{cases} \text{Strain hardening} \\ \text{Strain-rate hardening} \\ \text{Thermal softening} \end{cases} \left( \frac{1}{1 - (T/\Theta_m)} \right) \]  

(28)

where \( A + Be^n \), \( 1 + Cln\dot{\varepsilon}^\prime \), and \( 1 - (T/\Theta_m) \) describe the isotropic strain hardening, strain rate hardening and thermal softening of the metallic materials, respectively. In the J-C model, the strain hardening, strain rate hardening, and thermal softening are assumed to be independent phenomena and can be separated from each other. Thus, the unknown parameters, such as \( A, B, C, n, \) and \( m \), are easily calculated by fitting the stress-strain curve at different strain rates. \( A \) stands for the yield stress at a certain reference temperature and strain rate, \( B \) and \( C \) are the strain hardening and strain rate hardening coefficients, respectively, and \( n \) and \( m \) are the strain hardening exponent and thermal softening exponent, respectively. \( \dot{\varepsilon}^\prime \) is the dimensionless strain rate, and \( T^\prime \) is the homologous temperature. \( \dot{\varepsilon}^\prime \) and \( T^\prime \) can be expressed as:

\[ \dot{\varepsilon}^\prime = \frac{\dot{\varepsilon}}{\dot{\varepsilon}_0} \]  

(29)

\[ T^\prime = \frac{T - T_r}{T_m - T_r} \]  

(30)

where \( \dot{\varepsilon}^\prime \) is the strain rate and \( \dot{\varepsilon}_0 \) is the reference strain rate, respectively. \( T, T_r, \) and \( T_m \) are the absolute, reference, and melting temperatures, respectively. All the tensile tests in the current study were performed at room temperature; thus, the effect of thermal softening can be neglected, which leads to the simplification of Eq. (28) to

\[ \sigma = (A + Be^n)(1 + Cln\dot{\varepsilon}^\prime) \]  

(31)

The material constants \( A, B, C, \) and \( n \) were obtained by fitting the stress-strain curve acquired from experimentation as follows:

(a) Determine \( A, B, \) and \( n \)

Initially, when \( \dot{\varepsilon}_0 = \dot{\varepsilon} \), Eq. (31) will be reduced to

\[ \sigma = (A + Be^n) \]  

(32)

In the current study, \( \dot{\varepsilon}_0 \) was selected to be \( 0.001 \) s\(^{-1}\). Thus, the material parameters \( A, B, \) and \( n \) were easily calculated from quasi-static stress-strain results.

(b) Determine \( c \)

Once \( A, B, \) and \( n \) were successfully calculated, \( c \) was determined when \( \dot{\varepsilon} = 0 \), which leads to the conversion of Eq. (31) to

\[ \sigma = A(1 + Cln\dot{\varepsilon}^\prime) \]  

(33)
The material parameter $C$ was calculated from the stress-strain results obtained at the strain rates (i.e., $10^{-2}$, $10^{-1}$, 1, 1733, 3098, 3651, and 3919 s$^{-1}$) other than $\dot{\varepsilon}_0 = 0.001$ s$^{-1}$. The final values of $A$, $B$, $C$, and $n$ determined by fitting are listed in Table 4.

**Verification of original Johnson-Cook model**

The accuracy of original Johnson-Cook model to predict the flow behaviour of AA2060-T8 sheets at QSR, ISR, and HSR was verified through comparing the results obtained from original J-C model with these acquired from experimentation as shown in Fig. 13a and b. As depicted in these figures, there are huge variations between the predicted and the experimental results, notably, when the strain rate goes to HSR region, which means that the original J-C model cannot sufficiently describe the flow behavior of AA2060-T8 sheets at room temperature and wide range of strain rates. These huge variations were attributed to the four material parameters (i.e. $A$, $B$, $C$, and $n$), since they cannot display the flow behavior of AA2060-T8 sheets sufficiently, and ignore the change in strain rate hardening coefficient ($C$) especially at HSR region.

**Improved Johnson-Cook model**

The original J-C model was modified to consider the change of the strain rate hardening coefficient ($C$) in the HSR region. The calculation method used to determine $A$, $B$, and $n$ in the improved J-C model is identical to the method used in the original J-C model. However, in the improved J-C model, the authors proposed a new equation to consider the effects of both strains ($\varepsilon$) and strain rates ($\dot{\varepsilon}$) on the strain rate hardening coefficient ($C$), as expressed in Eq. (34).

### Table 4

Parameters of J-C model of AA2060-T8 sheet.

| Material parameter | $A$ (MPa) | $B$ (MPa) | $C$ | $n$ |
|--------------------|-----------|-----------|-----|-----|
| Value              | 490       | 285       | 0.018 | 0.13545 |

Fig. 13. Comparison between experimental flow behaviors of AA2060-T8 sheets and those predicted using both original and improved J-C at (a) QSR and ISR and (b) HSR, where: $\sigma_\varepsilon$ is the experimental flow stress, as well as, $\sigma_{OJC}$ and $\sigma_{IJC}$ are the flow stress predicted by original and improved J-C model.
\[ C = f(e, i^{inc}) \]  

In the proposed equation, the parameter \( C \) was expressed as a quadratic polynomial equation containing \( e \) and \( i^{inc} \) as variables because of the complicated interaction between \( C \) and these variables. The proposed formula can be expressed as follows:

\[ C = C_0 + C_1 e + C_2 e^2 + C_3 e i^{inc} + C_4 (i^{inc})^2 + C_5 (i^{inc})^3 \]

where \( C_0, C_1, C_2, C_3, \) and \( C_4 \) are the regression coefficients determined through the optimum regression methods. The values of these parameters are listed in Table 5.

**Verification of improved Johnson-Cook model**

The results of flow stresses obtained from the improved J-C model were compared with these achieved from experiments as depicted in Fig. 13a and b. As shown in these figure, a remarkable agreement between the results acquired from improved J-C model and experimentation in all strain rate conditions, which signifies that the improved J-C model can predict the flow behavior of AA2060-T8 sheets at room temperature and under a wide range of strain rates. These superb agreements are caused by considering the change of the strain rate hardening coefficient (\( C \)) under a wide range of strain rates, as well as link the relationship between \( C \) and both \( e \) and \( i^{inc} \).

Further validation was carried out to quantitatively measure the predictability of the proposed strategy is significant. Eq. (36) was used to calculate \( R \) as:

\[ R = \frac{\sum_{i=1}^{N} (E_i - E)(P_i' - P)}{\sqrt{\sum_{i=1}^{N} (E_i - E)^2 \sum_{i=1}^{N} (P_i' - P)^2}} \]

where \( E_i, E, P_i', P \) and \( N \) are the experimental flow stress, mean value of the experimental flow stresses, the flow stresses predicted by improved J-C model, mean value of the predicted flow stresses and the total number of points used in this investigation respectively. AARE and RMSE are unbiased parameters used to quantify the ability of the improved J-C model to predict the flow behavior exactly [43–46]. AARE and RMSE were calculated by Eqs. (37) and (38), where, the small amount of AARE means that the reliability of the improved J-C model is remarkable and vice versa [46].

\[ \text{AARE} \% = \frac{1}{N} \sum_{i=1}^{N} \left| \frac{E_i - P_i'}{E_i} \right| \times 100 \]  

\[ \text{RMSE} = \sqrt{\frac{1}{N} \sum_{i=1}^{N} (E_i - P_i')^2} \]

The last statistical parameter is NMBE, which used to quantify the mean bias in the predictions from improved J-C model, where, the negative and positive values of NMBE indicate under-prediction and over-prediction respectively [47]. NMBE was calculated by Eq. (39) as:

\[ \text{NMBE} \% = \frac{(1/N)\sum_{i=1}^{N} (E_i - P_i')}{(1/N)\sum_{i=1}^{N} P_i'} \times 100 \]

**Table 5**
The Parameters of improved J-C model of AA2060-T8 sheet.

| Material parameter | A (MPa) | B (MPa) | \( n \) | \( C_0 \) | \( C_1 \) | \( C_2 \) | \( C_3 \) | \( C_4 \) | \( C_5 \) |
|--------------------|---------|---------|---------|---------|---------|---------|---------|---------|---------|
| Value              | 490     | 285     | 0.13545 | 0.1726  | 0.3925  | -0.2867 | -0.02908| -0.02736| 0.00117 |

**Table 6**
The values of R, ARRE\% , RMSE, and NMBE\% at a wide range of strain rates.

| Statistical parameter | R    | ARRE\% | RMSE (MPa) | NMBE\% |
|-----------------------|------|--------|------------|--------|
| Value                 | 0.9885 | 4.235428 | 6.991238 | -5.298 |

![Fig. 14. The correlation between the experimental and the predicted flow stresses determined using (a) original and (b) improved J-C model.](image-url)
The correlations between the results predicted by original as well as improved J-C model and those obtained from experimentation at different strain rates are presented in Fig. 14a and b respectively. Furthermore, the values of R, ARRE%, RMSE, and NMBE% for improved J-C model are listed in Table 6. These value verified that the improved J-C model can predict the flow behavior of AA2060-T8 exactly at room temperature and a wide range of strain rates.

Conclusions

Based on the results achieved in the current study, the main conclusions can be deduced as follows:

- At QSR and ISR deformation, the yield (YS) and ultimate tensile (UTS) strengths of AA2060-T8 sheets increased with increasing strain rate. In contrast, the elongation to fracture (ELf) decreased with increasing strain rate. Furthermore, these tensile properties was clearly dependent on the fibre orientation under QSR and ISR, which signifies that the tensile properties of the AA2060-T8 sheets exhibit a serious degree of in-plane anisotropy. The fracture modes of AA2060-T8 varied from ductile to brittle fracture with respect to fibre orientation and strain rate, which is in line with the results obtained from experimentation.

- The work hardening behavior observed in HSR deformation is more prominent than that detected in QSR and ISR deformation. Thus, the YS, flow stress, and UTS at HSR were higher than their counterparts at QSR and ISR, which generally attributed to the increased strain rates. Nevertheless, the strain-hardening rate observed in HSR deformation was lower than that in QSR and ISR deformation which attributed to the competition between the strain hardening and thermal softening as a result of increasing the adiabatic temperature with increasing the strain rate.

- The ductility of AA2060-T8 sheets was improved under HSR deformation because of the adiabatic softening with increasing the strain rate and the inertia effect which may contribute to diffuse necking, slow down the necking development and delay the onset of fracture. Thus, the fracture behavior of AA2060 undergoes ductile fracture under HSR deformation. These results can efficiently be used to develop a new manufacturing route based on IHF to manufacture thin-walled-complex shape components from AA2060-T8 sheets at room temperature.

- The Johnson–Cook constitutive model was modified to consider the effects of both strain and strain rates on the strain hardening coefficient to predict the dynamic flow behavior of AA2060-T8 over a wide range of strain rates. A remarkable accordance between the results acquired from the improved Johnson–Cook model and from experimental work under all strain rate conditions was observed. This result means that the improved Johnson–Cook model can predict the flow behavior of AA2060-T8 sheets at room temperature and over a wide range of strain rates.

Conflict of interest

The authors have declared no conflict of interest.

Compliance with Ethics Requirements

This article does not contain any studies with human or animal subjects.

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