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Dynamic recrystallization of a wrought magnesium alloy: Grain size and texture maps and their application for mechanical behavior predictions

Yuan Li, Peijun Hou, Zhenggang Wu, Zhili Feng, Yang Ren, Hahn Choo

**HIGHLIGHTS**

- Processing-grain size-texture maps were established as a function of hot-working conditions for AZ31B Mg alloy.
- The grain size map showed that the recrystallized grain size decreased with the increase in Zener-Hollomon parameter.
- The processing-texture map showed that the initial texture was altered significantly during the recrystallization process.
- The Schmid-factor map was established to identify a dominant deformation mode during a subsequent deformation.
- Yield-strength maps of the Mg alloy, hot worked at a wide range of temperatures and strain rates, were predicted.

**GRAPHICAL ABSTRACT**

**ABSTRACT**

Effect of dynamic recrystallization (DRX) on the grain refinement and texture modification was studied by conducting a series of hot compression on AZ31B Mg alloy. Processing-grain size-texture maps were established as a function of temperature, strain rate, strain, and the Zener-Hollomon parameter (Z). Moreover, influence of simultaneous changes in the grain size and texture on tensile yield strength and ductility of the hot-worked Mg alloy was studied using the grain size and texture maps established. The processing-grain size map showed that the DRX grain size decreased with the increase in Z, also revealing various characteristics ranging from a grain growth, grain refinement, to a bimodal distribution of ultrafine grains and partially recrystallized grains. The effect of twinning on the grain refinement was also evident at high Z conditions. The processing-texture map revealed that the initial fiber texture was altered significantly to a shear, off-normal, or extension-twin texture with the increase in Z. The Schmid-factor maps were calculated to identify a dominant deformation mode during a subsequent tensile deformation of the hot-worked samples. Finally, a corresponding Hall-Petch relationship for each dominant deformation mode was used to establish tensile yield-strength maps, which agree well with the measured data.

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1. Introduction

Magnesium alloys have a great potential for broad structural applications in the transportation industry because of its high strength to mass ratio and corresponding improvements in fuel efficiency [1–4]. However, due to a strong plastic anisotropy and the insufficient number of slip systems, the formability of Mg alloys is not adequate for an extensive ambient-temperature processes. Recent studies have shown that the plastic anisotropy and limited ductility at room temperature can be improved by grain refinement and weakening the (0002) fiber texture (i.e., a texture with the c-axis predominantly aligned along the normal direction of a wrought Mg alloy plate) [5–7]. Various thermo-mechanical processing methods; such as uniaxial hot compression [8,9], hot tension [10,11], rolling [12,13], torsion [14,15], extrusion [16,17], equal channel angular processing (ECAP) [18–21], and friction stir processing (FSP) [22–25]; have been used to tailor the microstructure and texture of Mg alloys.

During a thermo-mechanical processing, a significant grain refinement could occur due to a severe plastic deformation and dynamic recrystallization (DRX) at elevated temperatures [18–26]. Also, Zener-Hollomon parameter (Z), a function of the strain rate and temperature, is often used to characterize the grain refinement and dominant deformation mechanisms related to the thermo-mechanical inputs imposed during the processing:

\[ Z = \varepsilon \exp \left( \frac{Q}{RT} \right) \]

where \( \varepsilon \) is the strain rate, \( Q \) is the activation energy, \( T \) is absolute temperature, and \( R \) is the gas constant. The dynamically recrystallized (DRXed) grain size generally decreases with the decrease in the processing temperature due to the low mobility of grain boundaries [8–11,27–30]. For example, Kim et al. [19] processed a pre-extruded Mg alloy using the ECAP at 548 K with an extrusion speed of 4 mm/s, and the grain size was refined from 24.4 μm to 15.8 μm after a single pass and to 8.7 μm after eight passes. Also, Woo et al. [25] applied the FSP on a rolled Mg alloy, with the tool rotation speed of 600 rpm and travelling speed of 0.98 mm/s, and refined the grain size from 50 μm to 17 μm. With the grain refinement, an increase in the tensile ductility is often observed in Mg alloys [7,31–33]. For example, Chino et al. [7] showed that the tensile elongation of AZ31B Mg alloy at 523 K increases from 30% to 93% with the decrease in the grain size from 30.1 μm to 6.3 μm. In addition, Mohan et al. [33] obtained ultrafine grain size (0.8 μm) in the AZ31B by using two passes of FSP (first, 700 rpm, 3.4 mm/s and then 500 rpm, 5.9 mm/s), which facilitated a superplastic tensile elongation of 257% at 483 K and a strain rate of 1 × 10⁻⁴ s⁻¹.

Moreover, the crystallographic texture is often altered significantly during a thermo-mechanical processing [15,19,21,34–38]. For example, Agnew et al. [34] showed that when an as-extruded AZ31B was ECAPed at 473 K with a speed of 25 mm/min, the grain size was refined from 60 μm to 49 μm and the initial fiber texture changed to a texture with the basal plane tilted 45° from the extrusion axis. Yu et al. [35] studied the texture evolution during a series of FSP along RD (rolling direction) of a hot-rolled AZ31B plate as a function of the Z parameter. The initial strong fiber texture of (0001) // ND (normal direction) changed to a shear texture of <0001> // RD (i.e., the FSP direction) at a low Z processing condition (e.g., 1200 rpm, 0.25 mm/s) and to a strong off-normal texture along 40° toward the RD under a high Z condition (e.g., 600 rpm, 1.1 mm/s). The deformation and recrystallization textures contribute to the overall texture modification and, in general, the orientation of recrystallized grains is similar to that of the deformed parent grains [11,39–42]. Yi et al. [11] studied the recrystallization of AZ31B under tension at 473 K and 523 K and found that the textures of the recrystallized grains at the grain boundaries and the parent grains exhibited minor intensity differences. Similarly, Cotram et al. [2] investigated the DRX texture of extruded pure Mg–Y alloys, processed by a channel die compression at temperatures ranging from 473 K to 723 K and a strain rate of 1 × 10⁻³ s⁻¹, and also showed that the texture of DRXed Mg–Y grains followed that of the deformed parent grains. Moreover, Chen et al. [39] carried out uniaxial compression tests at 523 K, 0.001 s⁻¹ along ND and TD (transverse direction) of a rolled AZ31B alloy to study the effect of twinning on the DRX texture. It was concluded that the DRXed grains retained the orientations of the twins formed during the hot compression.

In terms of the development of the deformation texture in Mg alloys, the deformation twins produce the most dramatic texture modifications. The extension twins form under a tension along the c-axis or a compression perpendicular to the c-axis, then propagate rapidly, and rotate the basal plane by about 86° [43–46]. In comparison, the contraction twins form under compression along c-axis and rotate the basal plane by about 56°. However, the overall changes in the texture due to the contraction twins are not as significant [43–45,47,48] partly because a small volume fraction of contraction twins can accommodate relatively large shear strains [45,49]. On the other hand, the dislocation slip is not effective in modifying the texture. For example, during the uniaxial deformation of a Mg alloy, the activation of basal, prismatic, or pyramidal slip at elevated temperatures appears to make little changes on the texture [6,40–42,50]. The implication of the texture modification on the mechanical behavior of Mg alloys has also been studied extensively [19,34,35,51]. For example, Agnew et al. [34] reported that the ECAPed specimen, with the basal plane tilted 45° from the extrusion axis, exhibits a lower yield stress and a higher ductility due to the activation of the basal slip from the modified texture. Yu et al. [35] studied the tensile properties of the texture-modified FSP samples and also showed that the tensile yield stress along the RD ((// FSP direction) decreased from about 150 MPa of the as-received hot-rolled sample to about 50 MPa for the FSP sample, while the elongation improved from 15% to about 40%.

The combined effect of grain refinement and texture modification was also studied [19, Kim et al. [19] studied the Hall-Petch (H–P) relationship of an ECAPed AZ61 alloy and found that the H–P relationship has a negative slope indicating a potential texture effect. Wang et al. [52] showed that the initial texture, which governs the dominant yielding mechanism, has a significant influence on the H–P relationship of a Mg alloy in that the prismatic slip has a higher H–P coefficient than the extension twin and basal slip. Yu et al. [53] recently reviewed the factors that influence the H–P coefficient, such as texture, grain size, temperature, and strain, and showed that texture affects the H–P coefficient significantly.

Therefore, it is of technical and fundamental importance to understand the effect of the DRX on simultaneous changes in grain size and texture of Mg alloys during a thermo-mechanical processing and their combined influence on the mechanical properties. In particular, even though the effect of thermo-mechanical processing conditions on the DRX and the resulting grain-size refinement and texture change has been studied extensively for Mg alloys, a systematic assessment of a combined processing–grain size–texture maps, covering a wide range of Z parameter relevant to typical thermo-mechanical processing conditions, is quite limited. More importantly, based on these processing–microstructure maps, the implications of the simultaneous changes in the grain size and texture on subsequent mechanical behavior of DRXed Mg alloys could be assessed using the deformation–mode specific H–P relationships established recently. To this end, the main goals of the current study are to conduct a systematic investigation on the effect of key processing variables on the grain refinement and texture development in a Mg alloy and to establish a processing – grain size – texture maps that could help guide the assessment of subsequent mechanical performances.

In this paper, the grain refinement and texture development during a series of hot compression of an AZ31B Mg alloy will be presented as a function of strain rate, temperature, strain, and Z. First, the grain size refinement process will be presented as a function of applied strain (from...
10% to 50%) for a given Z value (ranging from $10^6$ to $10^{15}$ s$^{-1}$). The kinetics of the DRX process and the relationship between the DRX grain size and the Z parameter will be discussed. Then, a processing – grain size map, describing the relationship among the temperature, strain rate, and DRX grain size, will be established. Similarly, the texture development during the DRX process will be presented and the processing – texture map will be established. In addition, we will present the measured tensile behavior of several hot-compressed Mg alloy specimens with varying combinations of grain size and texture. Finally, the convoluted influence of the grain refinement and texture modification on the subsequent tensile yield behavior will be discussed by: (i) identifying a dominant deformation mechanism based on the Schmid factor map (calculated based on the texture map) for various potential deformation modes for three different loading orientations and (ii) calculating the yield strengths based on deformation-mode specific Hall-Petch relationships (based on the grain-size map). The combination of the deformation-mode-specific Schmid factor map and the Hall-Petch relationships allowed a prediction of tensile yield-strength maps for different loading directions, which agree well with the measured data. The current experimental approach to establish the grain size and texture maps and its application to the prediction of tensile yield strength can be implemented for other combinations of hot-working conditions and subsequent deformation or forming processes.

2. Experimental details

2.1. Material and hot compression test

The material used in this study is a commercial wrought AZ31B Mg alloy plate. The dimensions of the rolled plate was 500 × 100 × 6.5 mm. Cylindrical samples (6 mm in diameter and 6 mm in height) with the compressive loading direction (CLD) parallel to TD were prepared by electrical discharge machining (EDM). Fig. 1a. Throughout the paper, RD, TD, ND, CLD, and TLD refers to rolling, transverse, normal, compressive loading, and tensile loading directions, respectively.

The cylindrical samples were deformed by uniaxial hot compression along TD using the Gleeble system 3500. The samples were heated at 10 K/s to a target temperature and then soaked for 10 s, followed by hot compression to a set strain value under the displacement control. The cooling rate after the deformation was about 100 K/s in all cases. The hot compression tests were conducted at temperatures ranging from 523 K to 823 K, strain rates from 0.0001 s$^{-1}$ to 1 s$^{-1}$, and applied strains from 10% to 50% as summarized in Table 1. The corresponding Z values for the testing conditions range from $2.6 \times 10^6$ to $2.4 \times 10^{16}$ s$^{-1}$. Note that the activation energy used for the calculation of Z is 164 kJ/mol based on our previous study [29]. In addition to investigating the effects of T and $\varepsilon$, the experimental matrix was designed to provide several cases of isoZ conditions where varying combinations of T and $\varepsilon$ result in the same Z value.

2.2. Metallography and high-energy synchrotron x-ray diffraction

The as-received and deformed samples were polished and etched for about 3 s using a mixture of 0.6 g picric acid, 3 ml acetic acid, 6 ml water, and 20 ml ethanol. Light optical microscopy was used to observe the microstructure, and the grain size distribution was analyzed using ImageJ software following ASTM E112 standard [54]. The initial grain size of the as-received sample is 16 (± 7) μm, Fig. 1d. All average grain sizes reported are area-fraction-based average values unless noted otherwise.

The crystallographic texture was measured for the as-received and deformed samples using high-energy synchrotron x-ray diffraction (sxRD) at the beamline 11-ID-C, Advanced Photon Source (APS) at Argonne National Laboratory, e.g., [55]. The wavelength used was 0.1173 Å, the incident x-ray beam was collimated to 600 × 600 μm, and the sample-to-detector distance was set at 1500 mm, which covered 22 diffraction peaks (from 10° to 123°) for the Mg alloy. Cylindrical texture samples (4 mm in diameter and 4 mm in height) were extracted from the hot-compressed samples with its axis parallel to the CLD, Fig. 1b. The texture data was collected by rotating the cylindrical sample around its axis from 0° to 180° with a step size of 30° to achieve a full pole coverage. The measured Debye-Scherrer rings were converted into diffraction patterns using Fit2D software. Then, the diffraction data analysis was conducted by the Rietveld refinement and E-WIMV method using the Materials Analysis Using Diffraction (MAUD) software [56]. The pole figure data was generated using MTEX, a quantitative texture analysis software [57]. The as-received Mg alloy plate has a strong (0002) fiber texture along ND, Fig. 1d.

2.3. Tensile test of hot worked samples

Additional set of cylindrical samples, with larger dimensions (10 mm in diameter and 15 mm in height), were hot compressed along the TD using the Gleeble system to an applied strain of 50% following the same procedures described earlier to extract tensile specimens for the investigation of the effect of the grain size refinement and texture modification on the subsequent mechanical behavior. The specific hot-compression conditions investigated, in terms of the Z parameter, are marked with an asterisk (*) in Table 1, where the deformation temperature ranges from 523 to 773 K and the strain rate ranges 0.001 to 0.1 s$^{-1}$. These five conditions were selected to cover a wide range of Z parameter from a low Z to high Z. But, also importantly, these conditions...
were selected to include samples with coarse to fine grain sizes and with qualitatively different initial textures based on the processing-grain size-texture maps established. Thereby, the combined effects of grain refinement and texture change due to the DRX on the subsequent tensile behavior could be studied effectively as a function of Z. Also, the specific hot-compression temperatures and strain rates were selected to ensure a relatively easier control of the Gleeble system for the production of high-quality DRX samples with adequate dimensions and integrity for the extraction of multiple tensile samples along different loading directions. On this set of samples, the optical microscopy and sXRD texture measurements were also conducted to ensure reproducibility of the DRX grain size and texture. Flat dog-bone tensile specimens with the TLD parallel to ND and RD were prepared using the EDM, Fig. 1c. The gauge section of the tensile sample was 3 mm long, 2 mm wide, and 1.5 mm thick. Tensile tests were carried out using Instron-1000, a high-resolution load frame, at a nominal strain rate of $10^{-3}$ s$^{-1}$. An extensometer was used to record the axial strain throughout the test.

3. Results

3.1. Grain refinement as a function of \( \varepsilon \) under various Z conditions

Fig. 2 presents the microstructure of AZ31B Mg alloy deformed by hot compression as a function of the Z parameter ranging from $10^6$ to $10^{16}$ s$^{-1}$. Also, for each Z condition, the changes in the microstructure with the applied strain from 10, 20, to 50% are shown. The deformation conditions for each Z are also noted using \( (T, \varepsilon) \). For example, \( (823, 0.0001) \) denotes the hot compression condition of 823 K and 0.0001 s$^{-1}$ for a Z value of $2.6 \times 10^6$ s$^{-1}$.

![Fig. 2](image-url)

**Fig. 2.** Microstructure development during the hot compression observed as a function of the applied strain under a wide range of Zener-Hollomon parameter (Z) condition. Note that \( (823, 0.0001) \) denotes the hot compression condition of 823 K and 0.0001 s$^{-1}$ for a Z value of $2.6 \times 10^6$ s$^{-1}$.
increased. At about $Z = 10^{12} \text{s}^{-1}$, the DRX initiates predominantly at the grain boundaries (e.g., at 10% strain) and the fraction of the DRX grains increases with the increase in the strain. At $Z = 10^{14} \text{s}^{-1}$, extension twins were first observed at 10% strain but the emergence of fine DRX grains mostly appear at the grain boundaries. Then, with the increase in the strain, the DRX fraction increases gradually. At $Z = 10^{15} \text{s}^{-1}$, thicker extension twins were clearly visible at 10% strain and the fine DRX grains were not observed at the grain boundaries unlike the case of $Z = 10^{14} \text{s}^{-1}$. When the applied strain was increased to 20%, a progress on the DRX was observed within the twin boundaries as well as at the grain boundaries. A significant grain size refinement occurred both in terms of the DRX grain size and the overall %DRX at 50% strain. The 50% strain samples (the top row of Fig. 2) clearly shows the effect of increasing Z on the DRX process, where the grain size decreases systematically by decreasing the deformation temperature and/or by increasing the strain rate. Moreover, for each Z condition, the evolution of microstructure with the applied strain illustrates the changes in the DRX mechanisms from the low Z to high Z cases.

Fig. 3 presents the quantitative descriptions of the microstructural evolutions observed in Fig. 2. Note that the “grain size” refers to an average size of all grains within the deformed microstructure, whereas the “DRX grain size” refers to an average size of the recrystallized grains only. When quantifying the recrystallized grain size and fraction in the hot-worked samples during the image analysis, the grains with sizes below the lower boundary of the initial grain size distribution (i.e., $9 \mu m$) are considered as the recrystallized grains. Fig. 3a shows the changes in the average grain size (of all grains) measured as a function of the applied strain at various Z conditions. The as-received grain size increased from 16.5 to $35.5 \mu m$ (about 115% increase) after the hot compression at $(Z = 3E6 \text{s}^{-1}, T = 823 \text{K}, \varepsilon = 0.0001 \text{s}^{-1})$. When the Z was increased by about two orders of magnitude to $(1E8, 773, 0.001)$ by decreasing the temperature and increasing the strain rate, the grain size increased by about 40% to $23.5 \mu m$. Further increase in the Z value to $(7E9, 723, 0.01)$ did not make much difference indicating a very limited grain growth. However, as it was observed in Fig. 2, the grain refinement was not prevalent either to reduce the overall average grain size under these deformation conditions.

For the three higher Z conditions $(Z > 1E11)$ presented in Fig. 3a, the average grain size decreased with the increase in the applied strain under a given Z, signifying the influence of the DRX grain refinement on the overall grain size distribution. Moreover, as the Z increased, the final average grain size became finer after 50% deformation. There are a few issues worth noting for these higher Z conditions. First, Fig. 3b shows that the average DRX grain size decreased with the increase in...
Z. For example, the average DRX grain sizes are about 6.0 and 2.6 μm for the cases of (6E11, 623, 0.01) and (9E14, 573, 1), respectively. Second, the increase in the strain, under a given Z condition, did not influence the average DRX grain size, Fig. 3b. However, the volume fraction of recrystallized grains (%DRX) increased with the applied strain, Fig. 3c, which rationalizes the overall refinement in the grain size with the strain observed in Fig. 3a. Third, the %DRX evolutions measured as a function of the strain (Fig. 3c) indicate that the kinetics of the DRX appears to change as the Z increases. For the (6E11, 623, 0.01) case, the %DRX increased rapidly to about 25%DRX at 10% strain and then saturated at about 20% strain, achieving maximum of only about 40%DRX with the strain level studied. In contrast, for the (9E14, 573, 1) case, the DRX process started slow with an incubation period until about 10% strain, and then progressed rapidly to a maximum of about 70% DRX, resulting in a higher volume fraction of very fine DRX grains. It should be noted that the plots to the measured data are for a visual guide. Therefore, either an interpolation or extrapolation beyond the measured data should be taken with a caution. For example, for the higher Z case (9E14), the %DRX may not saturate beyond 30% or even 50% applied strain as indicated by the line fitting. However, even though there are a limited number of strain levels investigated here, the qualitative difference between the lower Z and higher Z cases is apparent. The corresponding microstructures for the above two cases are shown in Fig. 3d, where the 6E11 case (left image) shows recrystallization at the grain boundaries at the early stages of the deformation, whereas the 9E14 case (right image) shows a significant DRX within the twinned daughters in addition to the original grain boundaries above 10% plastic deformation. The kinetics of the (9E13, 573, 0.1) case (Fig. 3c), where the extension twinning was prevalent but the DRX was mostly limited to the grain boundaries at the early stage of the deformation, falls in between the two cases discussed above. More discussions on the DRX kinetics will be presented in sections 4.1 and 4.2.

3.2. Grain size map as a function of T, ε, and Z

A processing-grain size map, representing the compilation of microstructures of the hot-compression samples deformed to 50% strain, is constructed as a function of temperature, strain rate, and Z, Fig. 4. The isoZ contour lines (dotted lines) are marked using the exponent value of the Z parameter (i.e., the number 12 on a dotted line denotes an isoZ contour line for Z = 1 × 10^{12} \text{s}^{-1}). First, the effect of temperature on the DRX grain refinement is evident. For example, at a given strain rate of 0.01 \text{s}^{-1} (or log ε = −2), the DRX grain size decreases gradually and relatively homogeneously with the decrease in the temperature from 723 to 523 K. Similarly, the effect of strain rate is also quite clear. For example, at given temperature of 623 K, the DRX grain size decreases systematically as the strain rate increases from 0.01 to 1 \text{s}^{-1}. Furthermore, the microstructures on a given isoZ contour line are quite comparable. In particular, the average DRX grain sizes on a given isoZ contour line, with varying combinations of temperatures and strain rates, are almost identical. The grain growth is evident below the isoZ line of 10^9 \text{s}^{-1}, whereas DRX grain refinements are clearly shown above the isoZ line of 10^{11} \text{s}^{-1}. In general, the DRX grain size decreases with the increase in Z. Above the isoZ line of 10^{14} \text{s}^{-1} (at the low T and high ε corner of the map), the DRX microstructures exhibit a bimodal distribution of the grain sizes, whereas very fine DRX grains are observed within partially recrystallized microstructures. The quantitative

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**Fig. 4.** Processing-grain size map representing the microstructure of hot-compressed AZ31B alloy deformed to 50% strain as a function of temperature (K) and strain rate (s^{-1}). The isoZ contour lines are also shown with a number denoting the exponent value (i.e., 12 denotes 1E12 s^{-1}).
relationship between the DRX grain size and $Z$ will be discussed in section 4.2.

3.3. Texture development as a function of $\varepsilon$ under various $Z$ conditions

Fig. 5 shows the development of the hot-compression texture as a function of the applied strain under various $Z$ conditions. The (0002) pole figures are presented for the samples compressed along TD under different combinations of temperature and strain rate. At a low $Z$ of $10^6 \text{s}^{-1}$, the initial (0002) fiber texture of the as-received plate remains strong until about 20% strain. Then, at 50% strain, it was observed that the (0002) poles rotated about RD forming a relatively weak and broad intensity distribution on the ND-TD plane. At the $Z$ of about $10^8 \text{s}^{-1}$, the initial (0002) basal texture started to diffuse along the ND-TD plane at a lower strain. For the $Z$ conditions between $10^9$ and $10^{12} \text{s}^{-1}$, the broad (0002) pole distributed along the ND-TD plane at about 20% strain evolved into a strong double-component, off-normal texture at about 60° from ND toward TD at 50% strain, e.g., see the (623, 0.01) case at 50% strain. The compression along TD would activate the extension twins with two dominant variants along the north and south poles at a lower temperature and/or at a higher strain rate. Indeed, the extension twin texture can be clearly observed at 10% strain at the $Z$ values higher than $10^{14} \text{s}^{-1}$. For the (573, 0.1) case, the extension twin texture appeared at TD at 10% strain, while the initial basal texture along ND weakened and spread along the ND-TD plane as observed at lower $Z$ conditions. At 20% strain, the extension twin formation is expected to complete and the ND component depleted further. At the same time, the 60° off-normal components appeared. Finally, at 50% strain, the texture consists of a combination of slightly weakened extension twin texture and the dominant 60° off-normal texture component. The (573, 1) case shows a very similar texture evolution as that of the (573, 0.1), but the extension twin texture component is much more dominant.

3.4. Texture map as a function of $T$, $\varepsilon$, and $Z$

The processing-texture map of the AZ31B Mg alloy plate, hot-compression deformed to 50% strain along TD, is presented using the (0002) pole figures as a function of temperature, strain rate, and $Z$ in Fig. 6. By following the iso$Z$ contour lines from the low $Z$ to high $Z$ conditions (i.e., from the lower left corner to the upper right corner of the map), the effect of $Z$ on the texture development can be observed clearly. The texture map has three regimes. Below $Z$ of $10^8 \text{s}^{-1}$, the deformation texture exhibited a diffuse distribution of (0002) intensity along the ND-TD plane. Between $Z$ of $10^9$ and $10^{13} \text{s}^{-1}$, the 60° off-ND component is the dominant texture. Above $Z$ of $10^{14} \text{s}^{-1}$, the texture can be represented as a combination of the 60° off-ND texture and the extension twin texture. With further increase in $Z$, the extension twin texture component gradually becomes stronger with a relative weakening of the off-ND component. As observed in the grain-size map, the pole figures on a given iso$Z$ contour line are quite comparable to each other.

3.5. Tensile behavior of hot-compressed specimens

The effect of changes in the grain size and texture in the hot-compressed specimens on the subsequent tensile behavior was studied by conducting a series of tension testing at ambient temperature as outlined in section 2.3. Fig. 7 shows the microstructure and texture of a subset of large hot-compression samples prepared for the tensile testing. The samples were pre-deformed to 50% compressive strain along
TD under a wide range of $Z$ conditions ranging from $1 \times 10^8$ to $2 \times 10^{15}$ s$^{-1}$. When compared to the results in Fig. 4 and Fig. 6, the microstructure and texture are quite reproducible.

Fig. 8 presents the stress-strain curves measured during the tensile loading (along ND) of the pre-deformed samples. First, the as-received specimen exhibits the initial yielding (at about 60 MPa) followed by a positive concave-up hardening behavior, which is typical of an initial extension twinning and subsequent dislocation slip during the c-axis tension along ND, e.g., [52]. The tensile behaviors of the samples pre-deformed at $Z$ values of $1 \times 10^8$ and $7 \times 10^9$ s$^{-1}$ are qualitatively similar.

Fig. 6. Processing-texture map representing the effect of temperature and strain rate on the texture development in terms of (0002) PFs after a hot-compression to 50% strain. The iso$Z$ contour lines are also shown with a number denoting the exponent value (i.e., 12 denotes $1E12$ s$^{-1}$).

Fig. 7. The microstructure and texture of a subset of large hot-compression samples deformed to 50% strain under the following ($Z$, $T$, $\varepsilon$) conditions: (a) $1E8$, 773, 0.001, (b) $7E9$, 723, 0.01, (c) $6E12$, 623, 0.1, (d) $9E13$, 573, 0.1, and (e) $2E15$, 523, 0.1. Tensile testing on these hot-worked samples was conducted at ambient temperature and the results are presented in Fig. 8.
to the as-received sample. Specifically, both the yield stress and the initial hardening behavior are quite comparable. However, the hardening rate is lower beyond about 7% strain and the tensile ductility is better especially for the sample prepared under a lower Z condition. The grain size refinement was not significant under these two low Z conditions but the texture changed significantly from the initial ND to off-ND components. Hence, the main cause of the changes in the hardening rate and ductility is likely due to the changes in the texture. The three samples pre-deformed at higher Z conditions exhibit concave-down, parabolic tensile stress-strain curves that are qualitatively different from the as-received and the two low-Z cases. When the sample was pre-deformed at a higher Z condition (i.e., from $6 \times 10^{12}$, $9 \times 10^{13}$, to $2 \times 10^{15}$ s$^{-1}$), the subsequent tensile yield strength increased from 110, 150, to 225 MPa and the ductility decreased from 27, 10, to 7%. In particular, the case of $Z = 6 \times 10^{12}$ s$^{-1}$ exhibits a combination of relatively high strength and ductility (Fig. 8) with its activation energy. Moreover, Eq. (8) allows a correlation between the recrystallized grain size and the Zener-Hollomon parameter as discussed in section 4.3.

4. Discussion

4.1. Kinetics of dynamic recrystallization

In this section, the effects of temperature and strain rate on the dynamic recrystallization kinetics will be discussed by identifying the activation energy and stress exponent and also by establishing a quantitative relationship between the DRX grain size ($D_{\text{DRX}}$) and the Z parameter. Fig. 9 summarizes the average $D_{\text{DRX}}$ measured (symbols) after the hot compression to 50% strain as a function of the temperature and strain rate. Fig. 9a shows that, under a given strain rate, the measured average $D_{\text{DRX}}$ decreases as the temperature decreases. Also, at a given temperature, the average $D_{\text{DRX}}$ decreases with the increase in the strain rate, Fig. 9b. The average $D_{\text{DRX}}$ becomes as small as 1.6 μm when deformed at 523 K and 1 s$^{-1}$ compared to the initial grain size of 16 μm.

The combined effect of $T$ and $\dot{\varepsilon}$ on $D_{\text{DRX}}$ can be evaluated as follows and the results are also shown in Fig. 9 (lines). First, according to Sandström [58], the $D_{\text{DRX}}$ can be estimated by:

$$D_{\text{DRX}} = \frac{M L G b^3 \rho_n^2}{\dot{\varepsilon}}$$  \hspace{1cm} (2)

where $M$ is the grain boundary mobility, $L$ is the mean slip distance of dislocations, $G$ is the shear modulus, $b$ is the Burgers vector, $\rho_n$ is the dislocation density, and $\dot{\varepsilon}$ is the strain rate. The grain boundary mobility, $M$, follows an Arrhenius type relationship:

$$M = M_0 \exp \left( - \frac{Q}{RT} \right)$$  \hspace{1cm} (3)

where $M_0$, is a material constant, $T$ is the absolute temperature, $R$ is the gas constant, and $Q$ is the activation energy for lattice diffusion.

According to Galiyev et al. [59], a constitutive relationship between $\dot{\varepsilon}$ and peak stress ($\sigma$) is:

$$\dot{\varepsilon} = A_1 \left( \frac{\sigma}{C} \right)^n \exp \left( - \frac{Q}{RT} \right)$$  \hspace{1cm} (4)

where $A_1$ is a material constant and $n$ is the stress exponent. Typically, $Q$ ranges from 140 to 230 kJ/mol and $n$ ranges from 4 to 7 for the hot deformation behavior of Mg alloys within a temperature range from 573 to 723 K [29,59,60]. Also, the relationship between stress and dislocation density can be written as [61]:

$$\sigma = \sigma_0 + A G b \sqrt{\rho_m}$$  \hspace{1cm} (5)

where $A$ is a material constant and $\sigma_0$ is the stress required to activate a dislocation in the absence of other dislocations. Since $\sigma_0$ is much smaller compared to $\sigma$, $\sigma$ can be approximated as $A G b \sqrt{\rho_m}$. Using Eqs. (3) and (5), Eq. (2) can be rewritten as:

$$D_{\text{DRX}} = \frac{M_0 L}{A^4 G^2 b \rho_m} \dot{\varepsilon} \exp \left( \frac{Q}{RT} \right)$$  \hspace{1cm} (6)

By combining Eq. (4) and (6), the relationship between the $D_{\text{DRX}}$, $T$, and $\dot{\varepsilon}$ can be obtained:

$$D_{\text{DRX}} = \frac{C M_0 L}{A^8 A_1^4 b} \exp \left( \frac{4}{n} - 1 \right) \frac{Q}{RT} \dot{\varepsilon} \left( \frac{4}{n} - 1 \right)$$  \hspace{1cm} (7)

Taking $\frac{C M_0 L}{A^8 A_1^4 b}$, as a constant $C$, Eq. (7) can be simplified as:

$$D_{\text{DRX}} = C \exp \left( \frac{4}{n} - 1 \right) \frac{Q}{RT} \dot{\varepsilon} \left( \frac{4}{n} - 1 \right)$$  \hspace{1cm} (8)

Eq. (8) can also be written as:

$$\ln D_{\text{DRX}} = C_1 + \left( \frac{4}{n} - 1 \right) \frac{Q}{RT} \dot{\varepsilon} \left( \frac{4}{n} - 1 \right) \ln \dot{\varepsilon}$$  \hspace{1cm} (9)

It should be noted that, combining Eq. (2)–(5), it was assumed that the activation energies in Eqs. (3) and (4) are the same. This results in a single expression where the recrystallized grain size can be described as a function of the strain rate and temperature with an apparent activation energy. Moreover, Eq. (8) allows a correlation between the recrystallized grain size and the Zener-Hollomon parameter as discussed in the next section.

The measured grain size data were fitted using Eq. (8) and, using a least squares fit analysis, the constants ($C$, $n$, and $Q$) are obtained. When all experimental data are fitted by varying all constants, the stress exponent, $n$ is 4.5, the activation energy, $Q$ is 151 kJ/mol, and $C$ is 83 with a sum of the squared residuals of about 0.31. The average $D_{\text{DRX}}$ calculated using Eq. (8) and these constants are also presented in Fig. 9 (lines), which shows a good agreement with the measured data under all conditions. Similarly, when $Q$ was fixed to 164 kJ/mol (a value
obtained from a previous study \[35,62\], the result showed that \( n \) remained at 4.5, \( C \) increased slightly to 86, and also maintained the goodness of fit with the residuals of about 0.35. It is worth noting that when the number-fraction-based average \( \text{DDRX} \) was used to obtain the constants, \( n \) and \( Q \) are 4.6 and 149 kJ/mol (with a sum of the squared residuals of 0.56), which are quite comparable.

4.2. Relationship between DDRX and Z

A phenomenological relationship between the hot-working conditions (i.e., \( Z \)) and the resulting average \( \text{DDRX} \) can be obtained by combining Eq. (1) and (8):

\[
\text{DDRX} = CZ^{(n-1)}
\]

Fig. 10 summarizes the average grain sizes presented as a function of \( Z \). The average \( \text{DDRX} \) decreases with the increase in \( Z \) \[28,29,59,63\]. In the present study, the slope of log (\( \text{DDRX} \)) vs. log (\( Z \)) is a function of \( n \) and is about \(-0.11\), which is consistent with the evaluations shown above leading to the Eq. (10).

It was shown that the evolution of \( \% \text{DRX} \) with the increase in the applied strain (Fig. 3c) is different between the relatively low-\( Z \) and the high-\( Z \) conditions. For example, the recrystallized grains were observed at the grain boundaries after 10\% strain for a \( Z \) value of \( 6 \times 10^{11} \text{ s}^{-1} \), but the DRX process exhibited an incubation period up to about 20\% strain for a \( Z \) value of \( 9 \times 10^{14} \text{ s}^{-1} \). While the dislocation slip is dominant at the low-\( Z \) conditions \[28–30,41\] where the prismatic and pyramidal slips can be active in addition to the basal slip at elevated temperatures \[64–66\], the extension twinning can be activated during the deformation at the high-\( Z \) conditions. The corresponding microstructure evolution (Fig. 2) also showed that, at the lower \( Z \) conditions, the grain boundary DRX is predominant and is observed at a lower applied strain; whereas, at the higher \( Z \) conditions, the extension twinning precedes the DRX occurring both at the twin and original grain boundaries at a relatively higher applied strain. The occurrence of extension twins at the onset of plastic deformation during the early stages of the TD compression is likely accompanied by insufficient dislocation slip systems available to accommodate the strain especially at the high \( Z \) conditions. Hence, there would not be enough dislocation strain energy stored for the DRX at the grain boundaries at a relatively low applied strains unlike the low \( Z \) cases. Subsequently, when sufficient extension twinning provides a significant grain rotation for the basal slip to be activated in the twinned daughters and also when the applied stress is high enough to activate the pyramidal slip system \[62,67\], the DRX would progress significantly at the grain boundaries and twin boundaries. Nonetheless, even though the dominant deformation mode changes and the kinetics of \( \% \text{DRX} \) evolution with the applied strain varies for different \( Z \) conditions, the relationships between the average \( \text{DDRX} \) vs. temperature and strain rate (Fig. 9) and the \( \text{DDRX} \) vs. \( Z \) (Fig. 10) can be expressed using a single set of \( n \) and \( Q \) for the entire \( Z \) range studied. According to Galiev et al. \[59\], the DRX mechanism could change from the discontinuous DRX (DDRX) to the continuous DRX (CDRX) with the increase in \( Z \). In the same work, it was also shown that the slope of the grain-size vs. \( Z \) curve changes correspondingly. However, in the current study, the microstructure evolution indicates that the DDRX process is still relevant at the high-\( Z \) conditions. For example, the sample deformed to 20\% strain at \( Z \) of \( 10^{15} \text{ s}^{-1} \) (Fig. 2 or Fig. 3d) clearly shows prevalence of
recrystallized grains at the original grain boundaries forming the neck-like structure. Moreover, the measured δDRX vs. Z curve (Fig. 10) does not show any slope changes even at the highest Z conditions, indicating that the DRX mechanisms may not have changed significantly under the current experimental conditions (such as the initial grain size, initial texture and hot compression direction, and the testing temperatures and strain rates). In fact, the studies on the DRX of Mg alloys showed that the DRX mechanisms at relatively higher Z conditions could still predominantly be DDRX [8] or a mixed combination of DDRX and CDRX [68,69].

Also shown in Fig. 10 are the average grain size of all grains (i.e., overall grain size shown in black square symbols), which is larger than the δDRX but has a very similar trend as a function of Z up to about 10^{14}–10^{15} s^{-1}. When Z is above 10^{15} s^{-1}, the overall grain size increases with the increase in Z. This is due to the reduction in the overall δDRX at high Z conditions as observed in Fig. 4, which leads to the increase in the overall grain size.

A few example cases of FSP, extrusion, and rolling are also presented in Fig. 10 for a comparison. Note that the activation energy of 164 kJ/mol was used for a consistency in the Z scale. Pérez-Prado et al. [13] studied the grain refinement of the AZ31B Mg alloy by hot rolling at 673 K and estimated strain rate of 40 s^{-1}. After about 80% reduction (and the strain of 7.4), the initial grain size of 38 μm was reduced to 3 μm. Since the material was fully recrystallized, the 3 μm grain size can be considered as the δDRX, which agrees well with the current results. Uematsu et al. [70] studied the grain refinement during a hot extrusion of AZ31B billets at an extrusion ratio of 67, working temperatures of 684 and 629 K, and the extrusion rates of 2.0 and 1.2 m/min (which correspond to the estimated strain rates of 1.4 and 0.8 s^{-1}, respectively [71]). The initial grain size of about 200–250 μm was refined to 2.9 and 2.1 μm, respectively, under these two extrusion conditions, which show reasonably good agreements with the current data. Finally, the grain refinement during the FSP of an AZ31B alloy was studied by Chang et al. [27] for the tool rotation speed from 180 to 1800 rpm with a fixed tool travel speed of 90 mm/min. The temperature was measured using the thermocouples and the strain rate was estimated using an empirical expression. The DRX grain sizes reported for the FSP case are comparable but somewhat larger than the current results. This might be due to a relatively slower cooling rate at the wake of the FSP without an active cooling system. From the measured thermal profile during the FSP, the cooling rate is about 5–8 K/s, whereas the initial cooling rate during the current hot compression study using the Gleeble system is at about 100 K/s. The effect of a relatively slow cooling rate on the grain growth at the wake of the friction stir welding of an AZ31B Mg alloy was also observed in other studies [62]. The effect of using a slightly different activation energy, Q (e.g., 151 vs. 164 kJ/mol) was negligible on the evaluation of other physical constants (i.e., n and C) and also on the goodness of the fits between the measured and calculated δDRX shown earlier in Figs. 9 and 10. However, the estimation of Z using the Q value of 151 kJ/mol would result in a slightly lower Z value compared to the cases with Q of 164 kJ/mol, which would, in turn, result in an overall shift of the Z plots.

4.3. Influence of texture modification and grain refinement on tensile behavior

4.3.1. Experimental evaluation

The implication of the changes in the texture and grain refinement in the hot-worked samples on the subsequent mechanical behavior is discussed in this section by experimentally evaluating the set of five samples tested for the ND-tension case (Fig. 8). Based on the measured texture distribution (Fig. 7), the maximum Schmid factors for the basal slip, prismatic slip, and extension twin are calculated for the case of tension along ND and presented as a function of the Z value used to prepare these samples in Fig. 11. For the tensile samples prepared at relatively low Z conditions (i.e., low-Z cases), the Schmid factor for the basal slip is the highest among the deformation modes considered for the ND-tension yielding behavior (Fig. 11) due to the off-ND texture distribution observed in the (0002) pole figures (Fig. 7a-c). For the sample prepared at Z = 6 × 10^{12} s^{-1}, the Schmid factor for the basal slip decreased slightly and that of the prismatic slip increased with the strengthening of the texture component along TD. Therefore, for the three low-Z cases, the basal slip is expected to be the most favorable yielding mode. For the tensile samples prepared at high Z conditions (i.e., high-Z cases), where strong TD texture component developed in the (0002) pole figure due to the extension twinning (Fig. 7d,e), the prismatic slip has the highest Schmid factor, indicating that the yielding will be predominantly facilitated by the prismatic slip during the tension along ND (Fig. 11).

Moreover, the effect of grain size on the yield stress and elongation during the ND-tension is evaluated for the same five samples. Fig. 12a presents the yield stress as a function of the average grain size (D^{1/2}). The Z conditions used to prepare the tensile samples are also marked for each data point. As expected, the yield stress increases with the decrease in the grain size (and with the increase in the Z values). More interestingly, the Hall-Petch slope for the low-Z cases (4 MPa mm^{−1/2}) is much smaller than that of the high-Z cases (22 MPa mm^{−1/2}). This is due to the fact that the Hall-Petch coefficient for the prismatic slip (dominant in the high-Z samples) is significantly higher than that of the basal slip (dominant in the low-Z samples) [52,53,72]. Fig. 12b shows the tensile elongation as a function of the grain size. The tensile elongation for the low-Z cases (22–27%) is generally much higher than that for the high-Z cases (7–10%). The decrease in the elongation observed for the higher-Z cases could be due to several reasons. First, in terms of the initial texture, the higher-Z cases are limited to the prismatic slip (which is likely followed by a pyramidal slip at higher applied stresses) since both basal slip and extension twinning are quite limited unlike the lower-Z cases according to the Schmid-factor calculations (Fig. 11). When the slip mode is limited almost exclusively to the prismatic slip, the tensile behavior typically exhibits a higher yield strength and an initial hardening rate but a lower ductility [52]. Moreover, at higher-Z conditions, it was shown that the DRX process is partially complete resulting in a mixture of fine DRX grains and relatively coarse grains. It is possible that a relatively higher dislocation density is retained in the hot-worked grains processed at higher Z, which, in turn, could result in a higher strength but reduced ductility. Furthermore, Fig. 12 illustrates the relative significance of the grain-size
refinement and texture change on the resulting mechanical behavior. The changes in texture (and corresponding changes in the deformation mode) have a significant influence on the tensile behavior, whereas the changes in the grain size have an added effect in the form of the Hall-Petch strengthening with a varying significance depending on the active deformation mode (e.g., basal vs. prismatic). More detailed discussion on this issue is presented in the following section. In short, depending on the processing condition, in terms of the Z values ranging from $1 \times 10^8$ to $2 \times 10^{15}$ s$^{-1}$, the tensile yield stress and the elongation could range between about 60 to 225 MPa and about 7 to 27%, respectively. Of the five samples tested, the best combination of tensile properties is displayed by the sample processed at $6 \times 10^{12}$ s$^{-1}$ (Fig. 7c) with the yield strength, tensile strength, and elongation of 110 MPa, 235 MPa, and 27%, respectively.

4.3.2. Yield strength prediction

4.3.2.1. Deformation-mode-specific Hall-Petch relationships. The combined effects of grain refinement and texture modification on the ambient-temperature tensile yield strength of the hot-compressed Mg alloy are evaluated more comprehensively by constructing the Schmid factor map and, then, the yield stress map for various loading scenarios based on the measured texture and grain-size maps. It is well known that the grain refinement increases the strength of the alloy following the Hall-Petch relationship [31,32]:

$$\sigma_y = \sigma_s + k_s D^{-1/2} \quad (11)$$

where $\sigma_s$ is the yield stress, $\sigma_p$ is the friction stress, $k_s$ is the strength coefficient, and D is the average grain size. According to Armstrong et al. [73], the corresponding shear stress relationship for a given deformation mode is:

$$\tau_y = \tau_{C_{RSS}} + k_s D^{-1/2} \quad (12)$$

where $\tau_y$ is the yield shear stress and $\tau_{C_{RSS}}$ is the critical resolved shear stress of the active deformation mode. The relationship between $\sigma_y$ and $\tau_y$ is:

$$\sigma_y = M \tau_y = M (\tau_{C_{RSS}} + k_s D^{-1/2}) \quad (13)$$

where M is the orientation factor, which is an inverse of the Schmid factor. Eq. (13) can be used to incorporate the effect of the changes in the grain size (D) and orientation (M) during a hot working on the resulting yield strength. The deformation-mode-specific $\tau_{C_{RSS}}$ and $k_s$ have been reported by Wang et al. in [52], where the Hall-Petch parameters for the basal slip, prismatic slip, and extension twin for the AZ31B Mg alloy have been experimentally evaluated (Table 2).

4.3.2.2. Calculation of Schmid factor map. The Schmid factors for the basal slip, prismatic slip, and extension twin are calculated, based on the strongest texture in the measured pole figures in the texture map (Fig. 6), for three different loading cases of tension along ND, RD, and TD of the samples previously hot compressed along TD. Then, the Schmid factor maps are constructed as a function of the processing conditions of temperature, strain rate, and Z parameter used for the hot compression, Fig. 13. The diameter of the circle in Fig. 13 is proportional to the Schmid factor value for each deformation mode.

For the ND-tension case (Fig. 13a), samples prepared with Z values below $10^{14}$ s$^{-1}$ exhibit the highest Schmid factor for the basal slip (i.e., the blue circle is the largest). The samples subjected to the Z values higher than $10^{15}$ s$^{-1}$ show the highest Schmid factor for the prismatic slip. This indicates that the dominant deformation mechanism during the yielding will be the basal and prismatic slip for the low-Z and high-Z cases, respectively. Moreover, based on the hierarchy of the Schmid factors for the different deformation modes, the extension twin could follow the basal slip for the very low-Z case during an early stage of plastic deformation; the prismatic slip could follow the initial basal slip in the mid-Z cases; but it would be exclusively prismatic slip for the high-Z cases. Certainly, in addition to the Schmid factor, the critical resolved shear stress is also important in predicting the yielding behavior and also the interactions of various deformation modes that could be activated during the subsequent plastic deformation. For the TD-tension case (Fig. 13b), the yielding would be almost exclusively facilitated by the prismatic slip. Finally, for the case of TD-tension (Fig. 13c), the deformation mode with the highest Schmid factor changes from the basal slip for the low-Z cases, to the basal slip and extension twin for the mid-Z cases, and to the extension twin for the high-

<p>| Table 2 |
|-----------------|-----------------|---------------|</p>
<table>
<thead>
<tr>
<th>Deformation mode</th>
<th>$\tau_{C_{RSS}}$ (MPa)</th>
<th>$k_s$ (MPa/(\text{s}^{1/2}))</th>
</tr>
</thead>
<tbody>
<tr>
<td>Basal slip</td>
<td>9.9</td>
<td>2.61</td>
</tr>
<tr>
<td>Prismatic slip</td>
<td>17.6</td>
<td>5.54</td>
</tr>
<tr>
<td>Extension twin</td>
<td>5.7</td>
<td>3.48</td>
</tr>
</tbody>
</table>
Fig. 13. Schmid factor maps of the AZ31 Mg alloy constructed as a function of temperature, strain rate, and Z parameters used for the hot compression along TD. The isoZ contour lines are also shown. The maximum Schmid factors for the basal slip, prismatic slip, and extension twin are calculated using the measured texture map (Fig. 6) for three subsequent loading cases of tension along (a) ND, (b) RD, and (c) TD at ambient temperature. The Schmid factors for different deformation modes are shown using different colors. The size of circle is proportional to the Schmid factor value. Also, specific Schmid factor values are provided above each data point.
Fig. 14. Yield stress maps of the AZ31B Mg alloy constructed as a function of temperature, strain rate, and Z parameters used for the hot compression along TD. The isoZ contour lines are also shown. The yield stress maps are generated using Eq. (11) and based on the Schmid factor maps (Fig. 13), grain size map (Fig. 4), and the deformation-mode specific Hall-Petch constants (Table 2). Three different loading cases are presented for the tension along (a) ND, (b) RD, and (c) TD at ambient temperature. The size of circle is proportional to the yield stress. The blue filled circle/number and the red hollow circle/number are the predicted and measured values, respectively.
Z cases. For a wide range of mid-Z conditions from $1 \times 10^{13}$ to $10^{14}$ s$^{-1}$, the Schmid factor values for the basal slip and extension twin are quite similar. Since their critical resolved shear stresses are also quite similar [52], the yielding could occur with both of the deformation modes activated.

4.3.2.3. Calculation of yield stress map. The activation of slip systems for the yielding is first assessed for each loading case by using the combination of the critical resolved shear stress and the Schmid factor. Then, the yield stress is predicted based on Eq. (13), utilizing the Schmid factor map for M (Fig. 13), the grain size map for D (Fig. 4), and the $\tau_{CRSS}$ and $k_\nu$ values for a specific deformation mode from Table 2. The yield stress maps are presented in Fig. 14 for three different loading cases, namely the tension along ND, RD, and TD of the samples prepared by the hot compression along TD. The diameter of the circle is proportional to the yield stress. Both predicted (blue filled circles/numbers) and measured (red hollow circles/numbers) data are presented for the ND- and RD-tension cases. Note that the measured RD-tension stress-strain curves are not presented here but they exhibit parabolic hardening with relatively high yield strengths and tensile elongations in the range of 10–18%. Also, note that the experimental validation for the TD-tension case is not available due to the difficulty in preparing the out-of-plane tension specimens (Fig. 1).

For the ND-tension case (Fig. 14a), the dominant deformation mode for the yielding would be the basal slip up to about $Z = 10^{14}$ s$^{-1}$, and, following the isoZ contour lines on the map, the tensile yield strength increases gradually with the increase in Z. Then, above $Z = 10^{14}$ s$^{-1}$, there is a sudden increase in the yield strength mainly due to the changes in the dominant slip mode from the basal to the prismatic slip. The slight decrease in the strength for the sample prepared at the highest Z condition (which is observed in all three loading cases) reflects the partial recrystallization and increase in the overall grain size observed in Figs. 4 and 10. Also, it is shown that the strength levels are comparable on a given isoZ contour line, which is reasonable considering that the grain size and texture are quite comparable on a given isoZ condition (Figs. 4 and 6). For example, on the isoZ contour line of $10^{13}$ s$^{-1}$, the yield strength ranges from 83, 80, to 78 MPa from a high-T/high-$\epsilon$ to a low-T/low-$\epsilon$ combination. The predicted yield stresses agree quite well with the experimental data across the Z range studied. However, the predicted yield stress of 95 MPa for the case of (9E13, 573, 0.1) underestimates the actual measured stress of 149 MPa. This is likely due to the fact that the yielding for the sample is assumed to be solely carried out by the basal slip, but in reality, there may be a measurable contribution from the prismatic slip due to its almost equally favorable Schmid factor. This is one of the limitations of the current approach in predicting the yield strength. Also, due to the distributions of texture components and the grain sizes, a development of constitutive relationships defining the role of texture in the Hall-Petch relationships is not straightforward in the current case. For example, a visco-plastic self-consistent modeling, which incorporates the deformation-mode dependent Hall-Petch relationships and texture distributions (e.g., [67]), would be a more comprehensive approach to predict the tensile behavior that includes the plastic flow and hardening behavior in addition to the yielding phenomenon. For the RD-tension case (Fig. 14b), the dominant yielding mode would be the prismatic slip across the entire Z range according to the Schmid factor map. The effect of grain size refinement is clearly visible with the gradual increase in the yield stress with the increase in Z. The predicted yield stresses match well with the measured values. Finally, on the TD-tension map (Fig. 14c), the yield stresses are relatively low. During the TD-tension, the dominant deformation mode at yielding makes a gradual transition from the basal slip for the low-Z samples to the extension twin for the high-Z samples. Since both deformation mechanisms have small $\tau_{CRSS}$ and $k_\nu$ (Table 2), the overall yield stress level remains relatively low with a weak dependency on the grain-size refinement with the increase in Z. The current approach to produce the Schmid factor and yield strength maps can be easily expanded for other mechanical testing scenarios. For example, the Schmid factor maps can also be constructed for other loading orientations and paths (e.g., compression) using the current texture map. The subsequent application of the Hall-Petch relationships for specific deformation modes identified from the Schmid factor map would allow the assessment of the yield behavior.

5. Conclusions

The grain refinement and texture development phenomena during a series of hot compression of a wrought AZ31B Mg alloy plate were investigated as a function of strain rate, temperature, and strain, covering a wide range of Zener-Hollomon parameter (Z) from about $1 \times 10^6$ to $1 \times 10^{16}$ s$^{-1}$. Based on the pole figures measured as a function of the hot-compression conditions, a processing – texture map was constructed. Similarly, the processing – grain size map was established based on the microstructure analysis of the hot-compression samples. Moreover, the processing – texture – grain size maps allowed the construction of the Schmid-factor maps and, finally, the yield stress maps for subsequent ambient-temperature tensile deformation of the hot-compressed samples along various loading orientations. The specific conclusions are as follows:

- The grain refinement process due to the dynamics recrystallization (DRX) during the hot compression was studied as a function of applied strain (from 10% to 50%) for a given Z value (ranging from $10^6$ to $10^{15}$ s$^{-1}$). A significant DRX was observed at the applied Z values above $10^{11}$ s$^{-1}$, where the %DRX increased with the amount of applied strain. However, the average DRX grain size ($D_{DRX}$) did not change with the increase in the strain.
- The kinetics of the DRX processes showed that when the DRX was preceded by the extension twinning at higher Z conditions, an incubation period was observed before the %DRX increased rapidly with the increase in the applied strain.
- The relationship among the average DRX grain size, temperature, and strain rate (Eq. 8) revealed that the stress component ($n$) is 4.5 and the pre-exponent constant ($C$) is 86 when the activation energy ($Q$) is 164 kJ/mol. The processing – grain size map showed that the $D_{DRX}$ decreased with the increase in Z following the relationship of: \( \log(D_{DRX}) = -0.11 \log(Z) + 1.9 \). Also, the $D_{DRX}$ was reasonably constant under a given Z condition with varying combinations of the temperature and the strain rate.
- The texture development during the hot compression along TD was studied and the processing – texture map was established using the measured (0002) pole figures. Below Z of $10^6$ s$^{-1}$, the texture exhibited a diffuse distribution of (0002) intensity along the ND-TD plane. Between Z of $10^7$ and $10^{13}$ s$^{-1}$, the 60° off-ND component was the dominant texture. Above Z of $10^{14}$ s$^{-1}$, a combination of the off-ND texture and the extension twin texture was prevalent. With a further increase in Z, the extension twin texture component became stronger with the weakening of the off-ND component.
- The subsequent tensile testing conducted along ND of the hot-compressed samples showed the effects of the changes in the texture and grain size. Specifically, the low-Z processed samples exhibited low yield strengths and a weaker grain-size dependency (i.e., a smaller Hall-Petch slope) with the dominant yielding mode of the basal slip according to their DRX texture. In contrast, the high-Z processed samples displayed higher yield strengths and a stronger grain-size sensitivity with the dominant prismatic slip.
- The yield stress maps were constructed to understand the convoluted influences of the grain refinement and texture modification occurred during the hot compression on the subsequent tensile yield behavior. First, the dominant deformation mechanism was identified using the Schmid factor maps generated for various potential deformation modes and loading directions using the processing – texture map.
The current approach to experimentally establish the grain-size, texture, and Schmid-factor maps and their application to the calculation of tensile yield strength, based on deformation-mode specific Hall-Petch relationship, could potentially allow the design and prediction of the mechanical behavior of hot-worked samples based on the thermo-mechanical processing conditions. This method can also be implemented for other combinations of hot-working conditions and subsequent deformation or forming processes. Finally, in addition to the yield behavior, the flow and hardening behavior of Mg alloys could potentially be investigated using the current approach when combined with a plasticity modeling scheme such as visco-plastic self-consistent modeling.

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

CRediT authorship contribution statement

Yuan Li: Conceptualization, Methodology, Validation, Formal analysis, Investigation, Data curation, Writing – original draft, Visualization.
Peijun Hou: Methodology, Investigation, Formal analysis, Visualization.
Zhenggang Wu: Methodology, Resources.
Zhili Feng: Methodology, Resources.
Yang Ren: Methodology, Resources.
Hahn Choo: Conceptualization, Methodology, Validation, Formal analysis, Investigation, Resources, Writing – review & editing, Visualization, Supervision, Project administration, Funding acquisition.

Declaration of Competing Interest
The authors declare that they have no known competing financial interests or personal relationships that could have appeared to influence the work reported in this paper.

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