Work hardening behavior of the extruded and equal-channel angularly pressed Mg–Li–Zn alloys under tensile and shear deformation modes

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ABSTRACT

The flow and work hardening behaviors of extruded and equal-channel angularly pressed (ECAPed) Mg–6Li–1Zn (LZ61) and Mg–12Li–1Zn (LZ121) alloys were studied by tensile and shear punch testing methods. It was shown that the Kocks–Mecking type plots for tensile and shear deformation of both alloys, exhibited similar work hardening (WH) stages in both extruded and ECAPed conditions. WH rates were found to be lower for the ECAPed materials, due to a reasonably uniform and well-refined microstructure. In the case of hcp LZ61 alloy, textural studies showed that the extruded fiber-type texture was replaced by a typical ECAP texture, in which basal planes rotated about 45° to the extrusion axis. This was found to be responsible for the lower tensile strength and higher shear strength in the ECAPed material, as compared to the extruded condition. For the bcc LZ121 alloy, it was observed that the grain refinement achieved after ECAP increases the strength and ductility in both tensile and shear deformation, compared with those of extruded condition. Stage II of the Kocks–Mecking plot in both shear and tensile deformation of LZ121 was eliminated most likely due to stacking fault energy improvement caused by higher Li content of the Mg lattice structure. The shear punch testing (SPT) method was found to yield the flow and WH curves similar to those obtained in tensile testing.

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1. Introduction

Magnesium alloys have attracted increasing attention from automotive and aerospace industries because of their low density, high specific strength and excellent damping capacity. Despite these advantages, the poor room temperature formability of Mg alloys, a consequence of the limited slip systems in the hexagonal close-packed (hcp) structure, restricts their applications [1,2]. Some of the possible means of addressing this deficiency are microstructural refinement [3,4], and use of alloying elements [5]. Among many possibilities, the promising severe plastic deformation technique of equal-channel angular pressing (ECAP) is capable of producing fine-grained microstructures [6]. The principles of the ECAP process have been reviewed by Valiev and Langdon [7]. Enhancement of cold formability of Mg alloys could also be achieved using Li as an alloying element. Mg–Li alloys, with Li contents between 5 and 11 wt%, exhibit a two-phase microstructure consisting of the hcp Mg-rich α and the body-centered cubic (bcc) Li-rich β phases. The single β-phase structure can exist for Li contents greater than 11 wt % [8]. A mixture of bcc and hcp phases is expected to change the deformation mechanisms and enhance the formability of the material, which is accompanied by a decrease in the strength [9].

Work hardening (WH) behavior, which influences strength and ductility, is one of the important considerations in evaluating the plastic deformation of materials [10]. Despite the great interest in the WH behavior of Mg alloys [11–13], only limited studies have been performed on Mg–Li alloys [14]. Wu and co-workers have investigated the plastic anisotropy and work hardening behavior of the cold-rolled dual-phase Mg–Li–Zn alloys at room [14] and high temperatures [15]. They used Kocks–Mecking type plots to illustrate different stages of work hardening and concluded that the anisotropic tensile behavior of the alloys may be related to the development of texture and microstructure during cold deformation.

Most of the WH studies have mainly focused on the investigation of tensile deformation behavior determined by the conventional tensile tests. There are, however, cases in which the material is only available as small thin test pieces such as those usually obtained by severe plastic deformation processes. In such circumstances, the miniature shear punch test (SPT) that is an easy-to-perform method [16] capable of evaluating stress–strain behavior of both cast [17,18] and wrought [19,20] magnesium alloys, can be advantageous. In this test, a flat-ended cylindrical punch is driven through a securely clamped sheet sample, punching a circular disc from it. By plotting shear stress against normalized displacement, mechanical properties such as shear yield stress (SYS), ultimate shear strength (USS) and...
shear elongation values can be obtained from the SPT data. To the best of authors’ knowledge, SPT has not previously been used for evaluating WH behavior of any alloy through determination of WH rates. It is therefore the aim of the current study to characterize and compare the tensile and shear WH behaviors of the fine-grained Mg–6Li–1Zn and Mg–12Li–1Zn alloys at room temperature. These were accomplished by measuring tensile and shear WH rates for alloys having various textures and grain sizes, obtained by conventional extrusion and by ECAP.

2. Experimental procedures

2.1. Materials and processing

This investigation involved two alloys with the nominal chemical compositions of Mg–6 wt% Li–1 wt% Zn and Mg–12 wt% Li–1 wt% Zn that are designated as LZ61 and LZ121, respectively. High-purity Mg (99.8 wt%), Zn (99.9 wt%), and an Mg–30 wt% Li master alloy were used to prepare the alloys. Melting was carried out in a graphite crucible placed in an electrical furnace under the protection of a covering flux. The melt was held at 750 °C for 20 min and mechanically stirred for 2 min, using a stainless steel rod before Li-containing master alloy was added successively every 10 min to it. An additional 20 min was allowed to ensure a homogeneous composition and to settle the oxides, before pouring the melt into a steel die preheated to 150 °C. Pouring was accomplished by a tilt-casting technique in order to minimize casting defects and the turbulence of the melt.

Extrusion was performed on the as-cast billets with an extrusion ratio of 11:1 at 300 °C. The ECAP process was conducted at 200 °C through route BC, in which each sample was rotated 90° around its longitudinal axis between the passages. This configuration leads to an imposed strain of about 1 on each passage through the die. Samples were sprayed with MoS2 lubricant and pressed at a speed of 1 mm/s for four passes using a solid die with channel angles of $\phi = 90°$ and $\psi = 20°$. The microstructure of the cross sections perpendicular to the pressing direction of the extruded and ECAPed billets was examined by optical microscopy. The metallographic samples of LZ61 and LZ121 alloys were etched with a solution of 5 g picric acid, 10 mL acetic acid, and 8% nitric acid and 92% ethanol, respectively. The intensity distributions of the (0002) and (1010) pole figures were measured by the Schultz reflection reference method from the plane perpendicular to the pressing direction for both extruded and ECAPed materials. The measurement was performed using Cu Kα radiation at 50 kV with the sample tilt angle ranging from 0° to 90°.

2.2. Mechanical testing

Miniature dog-bone tensile specimens, 4 mm long, 3 mm wide, and 2 mm thick, were prepared by electro-discharge wire-cut machining along the longitudinal direction of the processed materials. Tensile tests were carried out at room temperature with an initial strain rate of $1 \times 10^{-3} \text{ s}^{-1}$. Load–extension curves were obtained, from which the stress–strain curves, 0.2% yield stress, and tensile strength were automatically calculated by the machine where the data were acquired by a computer. Using this configuration, it was possible to control the load with the accuracy of 1/30,000 of the load cell nominal capacity of 20 kN, which was ± 0.66N, and to record the extension with the resolution of ± 0.001 mm. Three different samples were tested for each condition and the measured data were averaged. WH rates were obtained by differentiating the true stress–true strain data. This was accomplished by using a computer program, in which the derivatives were obtained by a five-point cubic spline numerical differentiation method.

Shear punch tests were performed using a SANTAM universal testing machine at an initial shear strain rate of $1.2 \times 10^{-3} \text{ s}^{-1}$. One-millimeter thick slices were cut from the extruded and ECAPed bars perpendicular to the pressing direction. These slices were mechanically ground to a thickness of about 0.7 mm and located in a fixture with a 3.175 mm diameter flat cylindrical punch and 3.225 mm diameter receiving-hole, the schematic of which is presented in Fig. 1. No lubricant was used between the specimens and dies. The applied load $P$ was measured as a function of punch displacement and the shear stress was calculated in MPa using the relationship:

$$\tau = \frac{P}{\pi dt}$$

where $P$ is the punch load in N, $t$ is the specimen thickness in mm and $d$ is the average of the punch and die diameters in mm. The shear strain ($\gamma$) was calculated from $\gamma = h/W$, where $h$ is the punch-displacement and $W$ is the die–punch clearance. Similar to tensile tests, three different samples were tested for each condition, and it was observed that the variation in the measured strength and elongation values was less than 2%.

3. Results and discussion

3.1. Microstructural observations

The microstructural evolution of the LZ61 alloy in the as-cast, extruded, and ECAPed conditions is depicted in Fig. 2a–c. As shown in Fig. 2a, the relatively large α grains with some β phase in the grain boundary areas can be observed in the as-cast microstructure. The extruded microstructure, shown in Fig. 2b, consists of the α-Mg matrix with some coarse deformed grains (15–30 μm) and fine recrystallized grains (2–8 μm). Some elongated β-Li constituents located at the α grain boundaries are also evident in the microstructure. Processing by ECAP leads to a more refined microstructure, so that a nearly uniform fine-grained structure with an average grain size of 6.3 μm forms, as exhibited in Fig. 2c. The grain refinement mechanism occurring during the ECAP process can be described as continuous dynamic recovery and recrystallization (CDRR), in which a combination of mechanical shearing and subsequent dynamic recovery, recrystallization and growth of grains and subgrains produce refined and equiaxed grains [21].

The micrographs showing the microstructure of the LZ121 alloy at different conditions are illustrated in Fig. 3a–c. The as-cast microstructure of LZ121 in Fig. 3a is indicative of coarse single phase β grains. In the authors’ previous work [6], the X-ray diffraction
analysis of the investigated alloys in the as-cast condition showed that with increasing the Li content from 6% to 12%, the α-phase peaks nearly vanish and that of the β-Mg phase increases, so that the XRD pattern of LZ121 mainly consists of the β-phase peaks. As can be seen in Fig. 3b, the inhomogeneous microstructure of the extruded condition is formed by a small fraction of large grains (115 μm) embedded in a matrix of relatively finer grains (21 μm). It seems that discontinuous dynamic recrystallization has completely occurred in the microstructure due to the high extrusion temperature (~0.66 Tm). The formation of some excessively coarse grains surrounded by dynamically recrystallized finer grains can be ascribed to the abnormal grain growth. This process, which is originated from the preferential growth of a few grains having some special growth advantage over their neighbors, is usually promoted in the strongly textured structures caused by rolling and extrusion [22]. Fig. 3c clearly shows the role of the ECAP process in the grain refinement of the LZ121 alloy by decreasing the average grain size from 30.3 to 6.1 μm. It is also evident that, after ECAP, a reasonably homogeneous fine-grained microstructure is obtained and the coarse-grained structure in the extruded condition entirely vanishes. As already pointed out, the microstructural evolution during the ECAP process could be the result of CDRR mechanism, in which the gradual transformation of low angle subgrain boundaries to high angle grain boundaries occurs by absorbing dislocations generated during successive passes of ECAP [23].

3.2. Textural evolution

To examine the textural evolution of the material during the extrusion and ECAP processes, the (0002) and (1010) pole figures of the extruded and ECAPed LZ61 are shown in Fig. 4. It is evident that the texture developed after extrusion, shown in Fig. 4a, has a
fiber-type character with the basal planes being parallel to the extrusion direction. In this state, the basal planes tend to align along the \(\langle 10\bar{1}0\rangle\) poles parallel to the extrusion axis, which is verified by the \(\langle 10\bar{1}0\rangle\) pole figure observed in Fig. 4a. After ECAP, however, most of basal planes depart from the extrusion direction, being located at about 45° to the extrusion axis, mainly due to the shearing parallel to the basal planes. Similar orientation of basal planes with respect to the pressing axis has been observed in different ECAPed Mg alloys [24,25].

Fig. 5 shows the \((110)\) and \((211)\) pole figures of the LZ121 alloy after extrusion and ECAP. In the extruded condition (Fig. 5a), the aggregation of the \((110)\) poles at the center of pole figure, forming a strong texture after extrusion, indicates that the \((110)\) planes are mostly located perpendicular to the extrusion axis. It is also evident that the \((211)\) poles are mainly inclined at about 25° to the extrusion direction, which develops a fiber-type texture after extrusion. Processing by ECAP encourages the formation of a new weakened texture with a characteristic three-point intensity and a randomized distribution in the \((110)\) and \((211)\) pole figures, respectively (Fig. 5b). Similar trend in the orientation of \((110)\) poles after ECAP has been observed in the bcc-structured IF-steel [26].

### 3.3. Tensile flow behavior

The tensile flow behavior of the extruded and ECAPed LZ61 samples is plotted in terms of true stress–strain curves in Fig. 6a. Concerning the yield stresses, the extruded material had a yield stress of 144 MPa and the ECAPed material had a yield stress of 137 MPa, though the corresponding grain size decreased from 9.6 to 6.3 μm. A ductility enhancement of about 48% can also be observed, where the elongation increases from 28.4% in the extruded condition to 42% in the ECAPed material. Such a decrease in strength together with the ductility enhancement in the ECAPed condition is attributed to the texture modifications occurring in the hcp crystal structure of Mg during ECAP. Primary slip occurs on the \((0001)\) basal planes in the Mg alloys at room temperature due to their low critical resolved shear stress (CRSS), as compared to those for non-basal slip systems [27]. For the extruded material, the basal planes tend to lie parallel to the extrusion axis (Fig. 4a), implying that primary slip would be difficult, and therefore, the yield stress increases. The rotation of basal planes (~45°) during the ECAP process (Fig. 4b), however, enhances the Schmid factor for the \((0001)\) basal planes, leading to a decrease in stress required for yielding of the ECAPed materials. The texture modification along with the grain refinement can also be the main reasons for the observed ductility enhancement [3]. However, the basal slip due to the texture modification cannot individually provide the necessary five independent slip systems for homogeneous deformation, according to the Von-Mises criterion. Accordingly, to sustain the large increase of tensile ductility, some non-basal slip planes have been activated due to a rotation of about 45° from the extrusion direction by ECAP. This argument is in agreement with the view of Agnew et al. in solid solution Mg–Li alloys [28].

Fig. 6b shows a Kocks–Mecking type plot of tensile WH rate \(\dot{\varepsilon} = \dot{\varepsilon} / \dot{\varepsilon}_t\) against net flow stress \(\sigma - \sigma_y\) for the LZ61 alloy in the extruded and ECAPed conditions. A nearly linear initial stage of WH is evident for both conditions, which may be related to the elastic–plastic transition, rather than dislocation mediated WH, as
proposed by Jain et al. [29]. No stage I hardening or ‘easy glide’, which depends strongly on the orientation of the crystal, can be observed in any of the tested conditions. The extruded condition exhibits an almost constant hardening behavior \( \dot{\theta}_II = \frac{d\sigma}{d\varepsilon} \approx \text{const.} \) associated with stage II of WH, with \( \theta_{II} \approx 880 \text{ MPa} \). The chief hardening mechanism has been related to the evolution of long-range stresses caused by dislocation pile ups at the grain boundaries [30]. It has also been suggested that the presence of stage II of WH might be due to the interactions of the dislocations in the primary slip system with those in an intersecting slip system [31]. Further increasing of stress is accompanied by a non-linear behavior that conforms to stage III of the Kocks–Mecking plot [32]. The WH plot of the ECAPed material, however, shows no stage II of WH, and the behavior is very similar to the well-known linear stage III of WH in fcc polycrystals. According to Rollett and Kocks [33], stage III of WH is characterized by a hardening rate which decreases monotonically with increasing flow stress arising from the parabolic hardening on the stress–strain curve.

It is worth describing the stress–strain curves and WH of the alloys by mathematical expressions as a widely used approach. Among many empirical equations, Voce equation [34], which is an alternative exponential stress–strain law compared to the common power law relation, has the advantage of showing an asymptotic saturation stress. Therefore, the WH rate in stage III can be well described by a differential form of the Voce equation in terms of net flow stress:

\[
\dot{\theta}_t = \dot{\theta}_0 \left( 1 - \frac{\sigma - \sigma_y}{\sigma_y} \right) \tag{2}
\]
where \( \theta_0^\infty \) is the WH limit extrapolated to \( \sigma = \sigma_y \), and \( \sigma_s \) is the saturation stress extrapolated to \( \eta_0=0 \). The respective values of \( \theta_0^\infty \) and \( \sigma_s \) obtained in this study are about 780 and 220 MPa respectively. The observed differences in WH behavior of the extruded and ECAPed conditions can be attributed to the grain size effects. A significant drop in WH rate of the ECAPed condition, compared to the extruded material, might be due to the grain refinement of about 34%, which leads to a higher contribution of grain boundary sliding \( \left( f_{\text{GBS}} \right) \) to the total deformation even at room temperature. This is because WH rate is proportional to the factor \( \left( 1 - f_{\text{GBS}} \right) \) [35]. The refined grain structure, in the ECAPed sample, is also responsible for the suppression of stage II and development of a linear stage III at the beginning of deformation via enforcement also responsible for the suppression of stage II and development of the activity of non-basal recovery process in stage III of WH, it should be mentioned that mention [36].

The grain-size dependent compatibility stress at grain boundaries is considered to be the main reason for this enhancement [36].

To characterize the influence of crystal structure on the mechanical properties, tensile flow behavior of the extruded and ECAPed bcc-structured LZ121 alloys is plotted in terms of true stress–strain curves in Fig. 7a. It can be observed that the yield stress increases from 84 MPa in the extruded condition to 111 MPa after ECAP. This is simply a consequence of the significant grain refinement from 30.3 to 6.1 \( \mu \)m that occurs after ECAP. It is believed that, as the Li content of the Mg-Li alloy increases, the crystal lattice axes ratio \( (c/a) \) of the hcp \( \alpha \) phase decreases so that slip between crystal planes become easier [37]. For Li contents greater than 11 wt%, a ductile bcc \( \beta \) phase structure develops, which has more independent slip systems than the hcp \( \alpha \) phase. Thus, if a relatively large stress concentration takes place in some areas during room temperature testing, the deformation can be accommodated by the \( \beta \) phase [38]. This results in the lower strength and higher ductility of the LZ121 alloy as compared to those of LZ61. A summary of the tensile and shear strength values of the investigated alloys in the extruded and ECAPed conditions is given in Table 1.

The Kocks–Mecking plots of the tensile WH rate against net flow stress for the LZ121 alloy in the extruded and ECAPed conditions are depicted in Fig. 7b. In both conditions, a linear elastic–plastic transition takes place at the beginning of plastic deformation, which moves toward a linear stage III as the deformation proceeds. Similar to the tensile WH behavior of LZ61 (Fig. 6b), the observed lower WH rates of the ECAPed condition in comparison with the extruded sample is due to the finer grain sizes achieved after ECAP. Since Li increasing tends to increase stacking-fault energy (SFE) in the Mg alloys [39], it is not surprising that only stage III of WH is discernible in the tensile WH plots of LZ121. It is reported that for materials with high SFE, stage III may limit the extent of stage II and even eliminate it as a separate stage [32]. For the LZ121 alloy, the values of \( \theta_0^\infty \) and \( \sigma_s \) in the modeled equation (Eq. (2)) are summarized in Table 2. It can be inferred that the above-mentioned effects are reflected in lower values of tensile saturation stress \( (\sigma_s) \), decreasing from 185 to 65 MPa for extruded alloys, and from 220 to 41 MPa for the ECAPed materials.

Tensile WH capacity \( (H^T) \) of a material can be defined in terms of ultimate tensile strength (UTS) and yield stress (YS) as [11]

\[
H^T = UTS - YS = \frac{UTS}{YS} - 1
\]

Table 1 depicts the tensile WH capacity of the LZ61 and LZ121 alloys in both extruded and ECAPed conditions. It is evident that processing by ECAP leads to a decrease in the tensile WH capacity of both alloys. \( H^T \) drops from 0.41 to 0.37 for LZ61 and from 0.34 to 0.07 for LZ121. It is believed that the WH capacity of an alloy is associated with its yield stress, which is further related to grain size in accordance with the Hall–Petch relationship [30,40]. Grain refinement would increase YS and decrease WH capacity. It also reduces the difference between flow resistance of the grain boundary and grain interior, which decreases WH capacity [41]. Therefore, the grain refinement caused by ECAP can be the main reason for decreasing WH capacity in both alloys. A comparison of tensile WH capacity of the LZ61 and LZ121 alloys at the same processing condition shows that it decreases with increasing Li content. This might be due to the fact that with increasing the amount of Li, the CRSS for non-basal slip diminishes in the hcp \( \alpha \) phase [42], activating the non-basal slip and facilitating the cross-slip between crystal planes, and thereby, decreasing WH rate [43]. Besides, the presence of the softer bcc-structured \( \beta \) phase with more independent slip systems, as compared to the hcp \( \alpha \) phase, could contribute to the lowering of WH rate. Accordingly, the WH capacities in the LZ121 alloy for both extruded and ECAPed conditions have lower values, compared to the LZ61 alloy.

3.4. Shear flow behavior

To study the shear flow behavior of the LZ61 alloy, the shear stress \( (\tau) \) was plotted against shear strain \( (\gamma) \) for both extruded and ECAPed conditions, as illustrated in Fig. 8a. It is evident that, analogous to the tensile stress–strain curves, after a linear elastic behavior the SPT curves deviate from linearity before they reach a maximum stress. The deviation point, obtained by plotting a tangent to the elastic part of the curve, is taken as the shear yield stress \( (\text{SYS} \ or \ \gamma_y) \) and the stress corresponding to the maximum point is referred to as the ultimate shear strength (USS). It can be seen in Table 1 that, in contrast to the tensile deformation, the SPT...
Table 1
Tensile and shear strengths (MPa) and WH capacity for the tensile ($H^T_s$) and shear ($H^{SPT}_s$) deformations.

<table>
<thead>
<tr>
<th>Materials</th>
<th>Extruded</th>
<th>ECAPed</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>YS</td>
<td>UTS</td>
</tr>
<tr>
<td>L761</td>
<td>144 ± 2.5</td>
<td>203 ± 3.6</td>
</tr>
<tr>
<td>L7121</td>
<td>84 ± 1.4</td>
<td>113 ± 2.1</td>
</tr>
</tbody>
</table>

Table 2
Extrapolated WH limit of stage III (MPa) for the tensile ($\theta^s_{II}$) and shear ($\theta^{SPT}_s$) deformations, and corresponding saturation stress (MPa) for both materials.

<table>
<thead>
<tr>
<th>Materials</th>
<th>Extruded</th>
<th>ECAPed</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>$\theta^s_{II}$</td>
<td>$\sigma_s$</td>
</tr>
<tr>
<td>L761</td>
<td>1150 ± 19</td>
<td>185 ± 3.2</td>
</tr>
<tr>
<td>L7121</td>
<td>1090 ± 18</td>
<td>65 ± 1.1</td>
</tr>
</tbody>
</table>

![Fig. 8](image-url)

Fig. 8. (a) Shear stress–strain curves, and (b) corresponding curves of shear WH rate vs. shear net flow stress for the extruded and ECAPed LZ61 alloy.

curve in ECAPed condition has higher values of SYS and USS (115 and 145 MPa), as compared to the extruded condition (94 and 122 MPa). This can be ascribed to the finer grain sizes achieved after ECA. Moreover, textural strengthening can partly enhance strength levels in shear deformation, the effect which was acting in opposite way in tensile deformation. In the extruded material, the basal planes are mostly aligned in the extrusion direction, and thus, the shearing during SPT most likely occurred on the basal planes under a lower yielding stress. In the case of ECAPed material, however, a decreased value of shear stress on the basal planes caused by the rotation of basal planes (~45° respect to the shearing direction) makes deformation more difficult, and thereby, the required stress for deformation increases.

Fig. 8b exhibits the shear WH rate ($\theta_s = \sigma_s/d\tau_s$) plotted against shear net flow stress ($\tau - \tau_s$) for the LZ61 alloy in the extruded and ECAPed conditions. It can be seen that the shear WH curves in both conditions show the same patterns as those obtained in tensile deformation (Fig. 6b). However, contrary to the tensile WH curves, the WH rate for the ECAPed sample is greater than that of the extruded sample. It seems that unfavorable orientation of basal planes respect to the shearing direction is responsible for the observed increase in the WH rate of the ECAPed material, the effect which offsets the influence of grain refinement. Similar to the tensile WH behavior, the shear stress dependence of the shear WH rate in stage III can be well modeled in the following form:

$$\theta_s = \theta^s_{II} \left( \frac{\tau - \tau_s}{\tau_s} \right)$$

(4)

where $\theta^s_{II}$ is the WH limit extrapolated to $\tau = \tau_s$ and $\tau_s$ is the saturation stress extrapolated to $\theta_s = 0$. The respective values of $\theta^s_{II}$ and $\tau_s$ are about 40 and 65 MPa respectively, as tabulated in Table 2. It is to be noted that the low values of $\theta_s$ observed in Fig. 8b are consistent with those of Les et al. [44], who investigated the shear WH rates of Al alloy using simple shear test.

The SPT curves of the LZ121 alloy for the extruded and ECAPed conditions are presented in Fig. 9a. It is clear that processing by ECA leads to higher values of SYS and USS as well as shear ductility. The texture-dependence of mechanical properties in the bcc-structured β phase is less pronounced compared to a hexagonally close-packed α phase, because of the higher lattice symmetry and more activated slip systems in cubic structures. Therefore, the significant grain refinement achieved after ECA could be the main reason for the observed strength and ductility enhancements of LZ121.

The Kocks–Mehcking plot of shear WH rate against net flow stress for the LZ121 alloy in the extruded and ECAPed conditions is shown in Fig. 9b. As expected, similar patterns to tensile deformation have been achieved in SPT of LZ121 in both conditions, as shown in Fig. 7b. This confirms that the SPT is a promising technique for evaluating WH behavior using small test pieces usually produced by severe plastic deformation processes such as ECA. Considering the linear stage III of WH in the both conditions, the values of $\theta^s_{II}$ and $\tau_s$ in the modified Eq. (4) are given in Table 2.

Tensile WH capacity equation (Eq. (3)) can be simply modified for evaluating shear WH capacity ($H^{SPT}_s$) in the SPT method by replacing UTS with USS and SYS with SYS as follows:

$$H^{SPT}_s = \frac{\text{USS} - \text{SYS}}{\text{SYS}}$$

(5)
This is based on the Von-Mises criterion for a state of pure shear of kinematically hardening materials, in which the relationship between tensile and shear stresses obeys $\sigma = \sqrt{3}\tau$. Employing this approach, the shear WH capacity of the alloys in both extruded and ECAPed samples is calculated and tabulated in Table 1. It is observed that, similar to tensile WH capacity, the grain refinement imposed by ECAP leads to the lowering of shear WH capacities in both alloys. The larger grain sizes provide more space to accommodate dislocations and result in the increased dislocation storage capacity, leading to enhanced shear capacities.

4. Summary and conclusions

The flow and work hardening behavior of the extruded and ECAPed Mg-6Li-12Zn (LZ61) and Mg-12Li-12Zn (LZ121) alloys were studied and the following conclusions are made:

1. The LZ61 alloy with an hcp lattice structure, showed higher tensile strength and lower ductility in the extruded condition, as compared to the ECAP condition. After ECAP, however, this trend was reversed mainly due to textural modification. The observed textural effect acts in an opposite way in shear deformation, due to the shearing nature of the SPT method, leading to the higher values of shear strength in ECAPed condition.

2. For the LZ121 alloy, it was observed that processing by ECAP increases the strength and ductility in both tensile and shear deformations, as compared to those of the extruded condition. Considering a weaker texture-dependence of mechanical properties in the bcc-structured LZ121 alloy, the significant grain refinement achieved after ECAP could be the main reason for the observed strength and ductility enhancements.

3. The study of work hardening showed that both tensile and shear WH rates of the hcp LZ61 correspond well to stages II and III of classical Kocks–Mecking plot in the extruded condition. However, the work hardening plot of the ECAPed samples exhibits only stage III work hardening, due to the refined microstructure caused by severe plastic deformation. For the bcc LZ121, no stage II was observed in both extruded and ECAPed conditions. Work hardening capacities of both alloys decreased after ECAP, mainly due to the higher density of grain boundaries provided by smaller grains.

4. It is demonstrated that shear punch testing (SPT) is capable of evaluating flow and WH behavior of the fine-grained Mg–Li–Zn alloys by measuring the WH rates. This can be particularly advantageous when the material is only available as small test pieces such as those usually obtained by severe plastic deformation processes. The validity of the SPT method was confirmed by obtaining the flow and WH curves similar to those obtained in tensile testing.

5. The relationships between different microstructures, textures, and deformation conditions are based on the initial microstructures and textures of the materials. Although the microstructural changes seem to be negligible, the textural evolution of the samples during both tests is expected to affect the WH behaviors. Unfortunately, textural studies of the shear-punched specimens in their very small shear zones are not feasible, and thus, it is not possible to elucidate textural evolution during SPT, to be compared with that in the tensile tests. Moreover, the tensile and SPT samples were selected from different positions in the processed materials and this might have an effect on some of the different trends observed in the tensile and SPT measurements.

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