Review Article

Strengthening Mechanisms of Magnesium-Lithium Based Alloys and Composites

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Mg-Li based alloys are widely applied in various engineering applications. The strength of these alloys is modified and enhanced by different strengthening mechanisms. The strengthening mechanisms of these alloys and their composites have been extensively studied during the past decades. Important mechanisms applied to strengthening the alloys include precipitation strengthening, solution strengthening, grain and subgrain strengthening, and dislocation density strengthening. Precipitation and solution strengthening mechanisms are strongly dependent on composition of the alloys and thermal treatment processes, whereas grain and subgrain and dislocation density strengthening mechanisms majorly depend on thermomechanical processing. In this paper, recent studies on conventional processes for the strengthening of Mg-Li based alloys are summarized as they are critical during the alloys design and processing. Main strengthening mechanisms are objectively reviewed, focusing on their advantages and drawbacks. These can contribute to enhancing, initiating, and improving future researches for alloys design and suitable processing selection.

1. Introduction

Mg-Li based alloys can be progressively used in the aerospace and aircraft structures as well as in ultraweight communications systems in the future due to their lightweight and strength [1–3]. The density of magnesium is approximately two-thirds of that of aluminium, one-quarter of zinc, and one-fifth of steel [4]. Lithium addition to magnesium bears an important effect on the crystal structure in particular lattice parameters. On increasing the lithium content in the alloy, the c/a axial ratio reduces from 1.624 for pure Mg to 1.607 for Mg-17 at.% Li, close to the solid-solubility phase. Magnesium alloyed with lithium above ~17 at.% Li (5.5 wt.%) makes the hexagonal-close-packed (HCP) structure. Mg-Li alloys change to a duplex structure mixed with both HCP and body-centered-cubic (BCC) phases simultaneously. Above ~31 at.% Li (11.5 wt.%), the alloys entirely change to BCC structure [1]. The brittle-to-ductile transition coupled with lithium’s lower density of 0.58 g cm$^{-3}$ presents substantial benefits of using Mg-Li based alloys for weight saving purposes [3, 5, 6]. Mg-Li system in Figure 1(a) signifies that the addition of lithium greatly enhances room temperature ductility of Mg-Li based alloys and their composites. The brittle-to-ductile transition temperature can be suppressed to liquid nitrogen temperature (Figure 1(b)) [1, 3, 6]. Thus, the addition of Li in Mg can lower the density of magnesium further and enhance the ductility and low temperature toughness simultaneously. Research evidences show that the applications of Mg-Li based alloys and composites are substantially hindered by their relatively low strengths and limited ductility at room temperature [4, 7, 8]. The magnitude of strength and ductility of Mg-Li based alloys and composites is majorly dependent on strengthening mechanisms which are controlled by alloy design and processing. Among the main challenges for materials scientists today who are working on Mg based alloys, understanding of strengthening mechanisms and their contributions is one of the most important prerequisites. This is because the alloys’ design, along with microstructure control, is strongly dependent on the application of one or more mechanical mechanisms [4, 8]. Strengthening mechanisms of Mg-Li based alloys and composites are generally ascribed to four well-known mechanisms. (i) Precipitation strengthening. This involves strengthening of grains due to formation of precipitates. (ii) Solution strengthening. This
40 Advances in Materials Science and Engineering

entails strengthening of the alloys phases because of dissolved atoms. (iii) Grain and subgrain strengthening. This strengthening mechanism occurs due to hampering of dislocation movement by the presence of grain and subgrain boundaries. (iv) Dislocation density strengthening. This strengthening of the magnesium-rich phase grains is attributed to impeding of dislocation movements formed within the alloys due to prior plastic deformation [4, 7, 8]. For many Mg-Li based composites, extruded hard phases, such as ceramic inclusions in the matrices, may strengthen the alloys further. This takes place through combining stiffness and hardness between the main pronounced phases present in the composites. In addition, the presence of ceramic inclusions may influence the precipitation in the matrices, hence indirectly initiating strengthening effects. This produces an important effect in determining the microstructure and mechanical properties of the Mg-Li based composites.

2. Modeling Strengthening Mechanisms in Mg-Li Based Alloys and Composites

2.1. Precipitation Strengthening. In binary systems, alloying above a specific concentration given by the phase diagram causes the formation of a second phase. A second phase can also be created through mechanical or thermal treatment methods. The second phase precipitates act as pinning points in a similar manner to solutes [14]. The precipitation strengthening effect in Mg-Li alloys depends on the size, morphology, distribution, and volume fraction of the precipitates and the nature of the interface between the precipitate and the matrix formed which can be coherent or incoherent. Dislocations cut through fine precipitates but cannot shear through larger incoherent precipitates. When the precipitates in an alloy cannot be cut across by dislocations, the Orowan increment of the Critical Resolved Shear Stress (CRSS) produced by the need for dislocations to bypass these obstacles is expressed by the equation below:

\[ \sigma_{ppt} = Z_0 M G b f^{1/2} R^{-1}. \]  

(1)

If the precipitates are cut by dislocations, the strengthening effect can be expressed as

\[ \sigma_{ppt} = Z_p M G K_p^{3/2} \left( f R^{-1/2} b^{-1} \right)^{1/2}, \]  

(2)

where \( Z_0 \) is a constant approximately 0.6, \( Z_p \) is a constant, \( M \) is the Taylor factor, \( G \) is the shear modulus, \( b \) is the magnitude of the Burgers vector, \( f \) is the particle volume fraction, \( R \) is the average radius of precipitates, and \( K_p \) is the parameter depending on the mechanism considered [15, 16]. Caili [17] found that alloying magnesium with lithium approximately 5–11 wt.% Li forms a dual \( \alpha \)- and \( \beta \)-phase eutectic lattice structure. A stable single \( \beta \)-phase Mg-Li alloy occurs when the Lithium content increases above 11 wt.%. The resultant alloy exhibits better formability and enhanced vibration resistance. Lithium in magnesium solid solution improves \( \alpha \)-phase toughness due to increased precipitation reactions [16]. The results from tensile simulation tests of \( \alpha \)- and \( \beta \)-phase boundaries show that dissolution of lithium in magnesium improves pure Mg-Li alloy interface binding capacity. This phenomenon promotes the bonding strength of Mg-Li interface in the alloy. A study of Mg-10Li-0.5Zn alloy by Wang et al. [9] found that the mechanical strength of the alloys steadily increases with aging time at room temperature. Maximum tensile strength achieved of 188 MPa, 180 MPa, and 173 MPa corresponds to the room temperature aging of 345.6 ks, 423 K aging of 21.6 ks, and 523 K aging of 10.8 ks, respectively. Figure 2 shows the sample hardness versus aging time for the solution-treated LZ101 alloy aged at room temperature, 423 K and 523 K, respectively. The precipitates dispersed in \( \alpha \)-phase effectively increase the mechanical strength of LZ101 alloy, exhibiting precipitation hardening at various aging temperatures [9, 18].

More recently, extensive studies have been carried out to study several effects of cold working and addition of alloying elements, aging and precipitation hardening. This is to enhance improvement in mechanical properties of Mg-Li based alloys and their composites. Commonly used elements are aluminum and zinc with the alloying combinations of Mg-Li-Al (LA-series), Mg-Li-Zn (LZ-series), and Mg-Li-Al-Zn (LAZ-series). Precipitation of \( \theta' \)-phase \([\text{MgLi}_2X, (X = \text{Al or Zn})]\) in the \( \beta \)-phase results in the hardening effect of
Mg-Li-(Al,Zn) alloys due to aging [3]. Xu et al. [19] found that the as-extruded LAM830-0.5Sr alloy exhibits an optimal elongation of 22.43%, which increases by approximately 43.97% compared with that of as-extruded LAM830 alloy. The as-extruded LAM830-0.75Sr alloy shows an increase in optimal tensile strength of 265.46 MPa by approximately 19.38% compared with the alloy without strontium. This indicates that the addition of rare earth, like strontium, can improve both strength and ductility. The studies on the microstructure evolution of Mg-14% Li-1% Al alloy found that the α-precipitates can coexist together with the ternary MgLiAl phase and embed in the β-phase grains homogeneously after ECAP processing. Thus the strength of alloys can be enhanced significantly [20]. Mendis et al. [21] reported the effect of lithium additions on the age hardening of Mg-2.4Zn-0.16Zr alloys. Their results show that pronounced precipitates form at the peak aging of 160°C. The precipitates are identified as intermetallic MgZn2 and lithium element is apt to be preferentially partitioned. Silver and calcium, when added to the alloy in trace quantities, enhance the overall age hardening ability. Investigation into the effect of aging treatments and microalloying on the microstructure by Sasaki et al. shows an increase in the peak hardness in Mg-Li-Zn-Al alloy [22, 23]. This is attributed to the synergistic precipitation effects and refinement of highly dense MgZn2 precipitates through heterogeneous nucleation. The previous research results show that T6 aging treatment with high peak hardness enhances the alloy creep property without degrading the room temperature strength. T6 is a two-step heat treatment process involving solution heating and artificial aging. T6 indicates that the heat resistance with superior room temperature strength can be obtained simultaneously due to precipitation hardening mechanism. Li et al. investigated the precipitation mechanisms in Mg-Gd-Nd-Zn-Zr alloys [24, 25]. Their results show that the precipitate phases known as β′′, β′, and β1 can form mainly through the significant partitioning of solute elements of Nd, Zn, and Gd. It is worth noting that the strengthening effect of β1 precipitates in the alloy is much stronger than β″ precipitates [24, 25]. This is evidence that even though this mechanism is of great substance in strengthening Mg-Li and composites, it does not exhaustively explain critical precipitation roles in the development of other recrystallized microstructures. Hence it is necessary to evaluate other alternative strengthening mechanisms and/or combine them together for strength enhancement.

2.2. Solution Strengthening. Solution strengthening is one of the most effective mechanisms used to improve the mechanical properties of Mg-Li alloy and its composites. This strengthening mechanism is scientifically explained from both elastic and electronic origins. The elastic origin is quantitatively justified based on the impurity-dislocation interaction atomic theory. The atomic size factor greatly contributes to the strengthening mechanism because it enhances Mg-Li alloy phases due to dissolved atoms [26, 27]. Solid solution strengthening (σss) occurs because the atoms in the solution alter and change lattice parameters and atomic binding forces of the alloys. In Mg-Li alloys and composites, the dissolved atoms effectively interact with dislocations and twinning, leading to the enhancement of strength. The atomic interactive intensity is determined by the size misfit parameter

$$\delta = \frac{\Delta a}{a}$$

and the modulus misfit parameter

$$\eta = \frac{\Delta G}{G}$$

where c is the atomic solute concentration, a is the lattice parameter, and G is the shear modulus. The strengthening effect of solid solution in Mg-Li based alloys and composites can be expressed by the following equations:

$$\sigma_{ss} = ZF \left(\frac{\Delta a}{a}\right) \left(\frac{\Delta G}{G}\right)$$

or

$$\sigma_{ss} = ZL \left(\frac{\Delta a}{a}\right)^{2/3} \left(\frac{\Delta G}{G}\right)^{2/3}$$

where ZF and ZL are constants and \(\delta\) lies between 1/20 and 1/16 [15].

A research study by Shi et al. evidently suggested that strengthening of polycrystalline Mg-Sn and Mg-Li binary alloys takes place through strengthening basal planes. By adjusting the grain size effects, its noted that 0.2% proof strength shows that the alloys obey a linear relationship law with C<sup>n</sup>, where c is the fraction of the atom and n = 1/2 or 2/3. However, this C<sup>n</sup> power-law is limited by relatively high temperature ranges due to the solution softening effect supposition [28]. Lithium in Mg-Li alloy decreases the c/a ratio of HCP magnesium lattice, leading to lower Critical Resolved

![Figure 2: Hardness versus aging time for the aged LZ101 alloy [9].](image-url)
Figure 3: Micrograph of Mg-8Li/SiC [10].

Shear Stress (CRSS) of the slip systems. This phenomenon initiates activation of more slip systems. Lithium solubility in HCP-structured magnesium is low, less than 5 wt.%, while it forms solid solution β-phase in a wide range of values. This is because large amounts of magnesium dominantly dissolve in lithium that possesses the BCC structure (Figure 1). Lithium effects on the deformability of Magnesium alloys occurs by replacing the less deformable HCP α-lattice with a BCC β-phase lattice which changes the alloys properties due to the appearance of β-phase [4, 29]. Maximum combination of the properties is realized within the range of binary (α + β)-phase alloys containing 6–10 wt.% Lithium. The optimized alloys dual parameter enhances the super plastic property of the alloys by raising their strength and ductility. Additional research on micro-hardness of these alloys has been done as shown in Table 1 [29]. From the research data, the plasticity of Mg-Li alloys becomes predominant with an increase in lithium content up to 16%. This also leads to an increase in the impact strength of Mg-Li alloy from approximately 7 [J/cm²] to 43 [J/cm²] according to Białobrzeski et al. [29]. Ye and Liu in their study found that Mg-Li binary alloys bearing the eutectic structures are composed of HCP and BCC solid solution dual phases. On dissolving in magnesium alloy, lithium initiates solution strengthening effect without forming Mg-Li precipitates [4]. Powder metallurgy method was employed in the study of microstructure of the composite reinforced with particles in Mg-8Li alloy having 7 vol.% of SiC. Finer grain sizes smaller than the size of the reinforcing particles within the alloy were used during the research study. From the micrograph in Figure 3, the microstructure exhibited two phases of bright HCP solid solution of lithium in α-phase and the dark BCC solution with the higher content of β-phase. The nonuniform distribution of particles with clusters of SiC particles initiated dispersive strengthening of Mg-Li alloy and its corresponding composites [1, 10].

In order to achieve optimum strengthening effects by solute solution mechanism on Mg-Li alloys and composites, solutes with higher shear modulus are used. This yields a higher local shear modulus in the alloys [30]. Alloying Mg-Li and its composites using elements having equilibrium lattice constants also enhances solution strengthening effect significantly. This is because greater difference in lattice parameters of the solutionized elements initiates higher local stress fields generated within the Mg-Li alloys lattice. These processes hinder dislocation propagation and finally lower the alloys’ plastic flow ability. This gives rise to increased yield strength proportional to the solute concentration in Mg-Li alloys [31, 32]. In a study by Chiang et al. [1], grain refining and strengthening of Mg-Li-Al-Zn-RE alloys by solid solution treatment followed by water quenching were carried out. A resultant microstructure of the supersaturated β-solid solutions was realized as shown in Figure 4. In-depth analysis of the supersaturated solution by XRD confirms the single phase as illustrated in Figure 5. The solution-treated alloys exhibit a unique significant increase in strength when compared with the as-extruded alloys as shown in Figure 6. Based on these results, it can be explained that the solution treatment strengthening mechanism effectively reinforces the Mg-Li alloys with adequate internal strain energy. Increased and enhanced strength and recrystallization with fine grains are achieved on further cooling and annealing [1]. Nevertheless, excess addition of solutes during solution strengthening may precipitate to form new phases. This is possible when the concentration of the solute exceeds a critical maximum point allowed by a specific binary phase system [30]. The critical solute concentration ultimately limits the degree of solid solution strengthening effect that can be achieved by other Mg-Li based alloys and composites. This limitation prompts the need to consider more alternative strengthening mechanisms.

2.3. Grain and Subgrain Strengthening. Importance of grain refinement in the strengthening of engineering alloys is extremely significant and beneficial. The yield strength of any material/alloy varies proportionally with the reciprocal square root of its measured grain size according to Hall-Petch theoretical equation:

\[
\sigma = \sigma_0 + Kd^{-1/2},
\]

where \(\sigma\) is the yield stress, \(\sigma_0\) is the yield stress of a single crystal, \(K\) is a constant, and \(d\) is the average grain size [16, 33]. The value of \(K\) is dependent on the number of slip systems and is greater for HCP metals than BCC and FCC metals. HCP metals exhibit higher strength sensitivity to the grain size. A study was done on the grain size effects on the strength of magnesium alloy AZ91 and aluminium alloy 5083. The results showed that the yield strength of magnesium alloy is lower than that of the aluminium alloy with grain size being larger than 2.2 \(\mu\)m. However, the yield strength of magnesium alloy increases steadily upon reducing the grain size to values less than 2.2 \(\mu\)m, as shown in Figure 7.

Table 1: Values of Vickers microhardness test measurements [29].

<table>
<thead>
<tr>
<th>Number</th>
<th>Examined alloy sample</th>
<th>Average microhardness</th>
</tr>
</thead>
<tbody>
<tr>
<td>1</td>
<td>MgLi-1.77</td>
<td>51.00</td>
</tr>
<tr>
<td>2</td>
<td>MgLi-2.1</td>
<td>59.07</td>
</tr>
<tr>
<td>3</td>
<td>MgLi-10.7</td>
<td>61.70</td>
</tr>
</tbody>
</table>

Number Examined alloys sample, Averagemicrohardness test measurements [29].
Advances in Materials Science and Engineering 5

Figure 4: Micrographs of solution-treated specimens: (a) LAZ1110, (b) LAZ1110+Be, (c) LAZ1110+Sc, and (d) LAZ1110+Be+Sc [1].

Figure 5: XRD patterns of solution-treated alloy specimens [1].

Figure 6: Tensile strengths solution specimens treated at 400°C and water quenched [1].

[4, 34]. Wang et al. carried out a study on the effects of calcium element on Mg-5Li-3Al alloy. Final results showed that marked grain refinement of as-cast alloys taking place on increasing calcium content from 0.5 wt.% to 1 wt.%. Mg-5Li-3Al-(0.5–1)Ca alloys exhibits enhanced mechanical properties mainly due to the grain refinement induced through the introduction of calcium atoms along with the formation of Al4Ca strengthening phase. It should be noted that higher amounts of calcium atoms (from 1 to 1.5 wt.%) lowered the tensile properties. This is attributed to the formation of (Mg, Al)2Ca eutectic phase along the grain boundaries due to the excess calcium atoms [35].

Lin et al. [34] investigated the effects of severe plastic deformation on Mg-Li alloys ECAP-ed and rolled the alloys after annealing. After ECAP and rolling, the tensile strength and strain elongation got improved. Further annealing of the alloys decreases the strength and increases strain elongation. The changes are due to the dislocation source-limited hardening and grain boundary strengthening mechanisms. Thermomechanical properties can be improved through reinforcing the alloys with the ceramic fibres and/or particles. An alloy with 8 wt.% of lithium consisting of dual α- and β-phases exhibits improved mechanical properties than single
Advances in Materials Science and Engineering

0.0 0.5 1.0 1.5 2.0 2.5
0
200
400
600
800
5083(H321)

\[ \sigma_y = 130 + 210\sigma^{-1/2} \] (MPa)

\[ \sigma_y = 230 + 63\sigma^{-1/2} \] (MPa)

\( \text{(Grain size)}^{-1/2} \) (\( \mu \text{m} \))^{-1/2}

Figure 7: Yield strength and grain sizes of AZ91 and 5083 alloys [4].

\( \beta \) -phase or \( \alpha \)-phase. The presence of interpenetrating (\( \alpha + \beta \)) matrix structure with the \( \delta \)-Al\(_3\)O\(_3\) fibres reinforces Mg-8Li. The reinforcements translate to increased composites strength as shown in Figure 8 [10]. Zhang et al. introduced long-period stacking ordered (LPSO) strips into Mg-Li alloys lattice matrix. Phase transition from \( (\text{Mg}_2\text{Zn})_2\text{Y} \) to \( \text{Mg}_{14.10}\text{Li-xSr} \) occurred with nonstoichiometrical compositions phases tolerating an appreciable amount degree of disorder. The disorder occurs at the Zn and Y sites in the Mg-8Li-6Y-2Zn alloy. The strength and plasticity of Mg-8Li-6Y-2Zn can be enhanced through refinement of grains and subgrains coupled with the formation of bent LPSO strips within the alloys [36, 37].

Nayak et al. [38] showed that biaxial rolling is effective in refining grains of Mg-3 wt.% Zn alloy. The yield strength and tensile strength increased by 65.32% and 77.34%, respectively, along with enhanced alloy toughness. Cheng et al. [11] in their research study developed a mechanism for strengthening new LAZ1151 Mg-Li alloy by use of cold rolling process (Figure 9). Strengthening effect is initiated through reduction of distortion of movement distance by refining the grain size (227 \( \mu \)m \( \rightarrow \) 147 \( \mu \)m). The \( \alpha \)-phase precipitations increased with more rolling reductions as a result of formation of metastable \( \beta \)-phase composites. When external energies were induced in to the lattice, the strain energy in the \( \beta \)-phase gets eliminated because the magnesium atoms form strong \( \alpha \)-phase composite.

Recent study research by Song et al. shows \( \beta \)-phase alloys of Mg(11-14)Li-Al possessing greater capacity for natural or artificial age strength hardening. The maximum hardening takes place only during aging process and correlates with the \( \theta \)-phase precipitation taking place at room temperature. The research showed that 1 wt.% aluminium cast alloy undergoes a change in hardness as a function of Reaction Time (RT) in Figure 10. The as-cast alloy at room temperature exhibits a reduced hardness magnitude on response to aging hardness than the sample measured after 2 days [12]. Wang et al. studied Al-Mg-Li alloys and showed that grain refinement occurred inhomogeneously with formation of ultrafine grains in bands. Resultant microstructures possessed approximately 70% high angle boundaries with the formed resultant grain structures being very stable at low temperatures. However, stability reduced at high annealing temperatures [27, 39]. From the experimental curves, it is concluded that the as-cast alloy hardens the same way as the as-quenched alloy and finally changes to age hardening alloy.

Figure 11 is a cast micrograph of three alloys (Mg-12Li, Mg-12Li-0.03Be-Al, and Mg-12Li-0.03Be-3Al) at low magnification power, revealing the grain structure of the \( \beta \)-solid solution along the grain boundaries. Grain refinement only takes place in the aluminium-containing alloys with the grain size of 1 wt.% Al and 3 wt.% Al alloy being reduced to 103 \( \mu \)m and 110 \( \mu \)m, respectively. The aluminium-free alloy had a large grain size of 190 \( \mu \)m [12, 39]. The addition of aluminium atoms produces coherent precipitates that increase the interaction between the lattice strain fields causing lattice distortion. Interactions between lattice strain fields initiated by grain refinement impede the induced dislocation motions. As a result, increase in hardness is realized with alloy of 3 wt.% Al. The hardness curve confirms that increased hardness occurred due to high aluminium content from the alloy micrographs. The high content initiates formation of large quantity of \( \theta \)-precipitates and interactions of lattice strain fields between the \( \theta \)-precipitates themselves [13, 40].

According to the Mg-Sr binary phase diagram in Mg14.10Li-\( \times \)Sr alloy, strengthening effect is attributed to formation of stable Mg-Sr compound. It is also possible because of effective grain strengthening effect for the formed \( \beta \)-phase owing to the crystallography matching relationship. The applied crystallography matching relationship is identified using the edge-to-edge matching model in different phases of Mg-Li alloys and composites [27, 40–43]. Al-3Ti-1C and Al-5Ti-1B master alloys are considered strong grain refiners for alloys with full BCC structure and HCP structures, respectively. The master alloys were used in refining Mg-Li-Al cast alloys in a study by Jiang et al. By increasing Al-3Ti-1C master alloy, the grain size of LA41 cast alloy was progressively lowered with the smallest grain size being 400 \( \mu \)m on introduction of 1.5% Al-3Ti-1C master alloy. For Al-5Ti-1B master alloy at 1.5%, the grain size of LA53 cast alloy reduced to 240 \( \mu \)m. This grain refinement mechanism of the cast alloys is fundamentally explained by heterogeneous nucleation effect between the particles of two compounds.
Advances in Materials Science and Engineering 7

Figure 9: Micrograph of LAZ1151: (a) 30% cold rolling, average grain size \( \sim 227 \mu m \), (b) 60% cold rolling, average grain size \( \sim 171 \mu m \), and (c) 90% cold rolling, average grain size \( \sim 147 \mu m \) [11].

Figure 10: Hardness of as-cast Mg-12Li and Mg-12Li-0.03Be-\( x \)Al \( (x = 1, 3) \) alloys with time [12].

formed (TiC and TiB\(_2\)) in the master alloys. The orientation relationship with regard to contact angle exists between the two compounds TiC and TiB\(_2\). Decrease in contact angle leads to decreased surface area of contact between the grain refiner and the matrix. This yields a geometrical effect in the alloys hence lowers the free energy barrier according to classical nucleation theory. This phenomenon contributes significantly during the grain refinement of the alloys [44, 45]. Higher density of grains and subgrains generally translates to enhanced yield stress and better strengthening effects of the Mg-Li alloys and composites. This is because of the emergence of increased subgrain boundary strength and intensity. The strength of these alloys varies reciprocally with the subgrain sizes which is analogous to the Hall-Petch theoretical relationship. Though grain and subgrain boundary strengthening is important, it is limited to strengthening very small grain sizes. This is attributed to breakdown point of most subgrains size occurring around 0.1\( \mu m \). Below 0.1\( \mu m \), decrease in yield strength of the alloys occurs and strengthening effect of the alloys is obviously reduced [46, 47].

2.4. Dislocation Density Strengthening. Ductile alloys increase in strength upon plastic deformation at temperatures lower than the melting point of the individual composites. Strain hardening in Mg-Li based alloys and composites takes place due to increased dislocation density and plastic deformation. This lowers the average distance between dislocations activating the dislocations to hinder the motion of each other. Strain hardening is defined as the percentage of cold work (% CW) and is calculated from the equation shown:

\[
\% CW = 100 \left( \frac{A_0 - A_d}{A_0} \right),
\]

where \( A_0 \) is the original cross-sectional area, \( A_d \) is the area after deformation, and % CW is a measure of degree of plastic deformation [48]. In their research study, Yang et al. [49] found that TEM examination shows high dislocation density at the grain interiors during extrusion of \( \alpha \)-based magnesium alloys due to Continuous Dynamic Recrystallization process. By introducing the theoretical Taylor equation:

\[
\sigma_{ds} = M \alpha G b \rho_{1/2},
\]

where \( M \) is Taylor factor, \( \alpha \) is a constant, \( G \) is the shear modulus, \( b \) is the Burgers vector, and \( \rho \) is the dislocation
density; the strength from a material dislocation density can be calculated. The calculated dislocation density in α-phase (1–101) planes of the extruded LA932 alloy is approximately $1.38 \pm 0.16 \times 10^{16}$ m$^{-2}$, estimated according the HRTEM image standards. From related studies the dislocation density calculated in (1–101) plane of α-phase in the as-cast magnesium alloy was found to be approximately $2.8 \times 10^{15}$ m$^{-2}$. Hence, it can be concluded that the dislocation density in (1–101) plane of α-phase increases due to extrusion process. From (6), high dislocation density translates to increased yield strength and enhanced mechanical properties of the alloys [13, 49, 50]. A study on deformation of Mg-Li-Al alloys shows decreased c-lattice parameter initiated by introducing Lithium atom. This initiates faster activation of nonbasal slip necessary for deforming alloy crystals due to high dislocation density. Dislocation of cross slip in the alloys becomes more pronounced at elevated temperatures [50–52]. Balogh et al. processed (Mg-5.5% Zn0.5% Zr) ZK60 alloy by ECAP for 8 equal passes and investigated its strength, dislocation density, and twin density. Both twin boundary density and the subgrain size did increase from the onset but gradually decreased after 2 ECAP passes. The dislocation density trend was notable on the decrease up to 8 equal passes owing to decreased mean free path for the dislocations. Effects of dislocation density and mean free path transformation of dislocations into boundaries as shown in Figure 12 significantly strengthened the alloys [13, 53].

Drozd et al. studied two sets of magnesium alloys (LA43 and LA45) at temperatures greater than 50°C. They found that the alloys characteristic stresses ($\sigma_{02}$ and $\sigma_{max}$) virtually...

Figure 11: Micrographs representing as-cast (a) Mg-12Li, (b) Mg-12Li-0.03Be-Al, and (c) Mg-12Li-0.03Be-3Al alloys [12].

Figure 12: Hardness, twin boundary density $\beta$, dislocation density $\rho$, and average area mean subgrain size $\langle x \rangle_{area}$ as a function of the number of ECAP passes [13].
decreased with increase in temperatures. Cross slip dislocations was pronounced at elevated temperatures and was the dominant strengthening mechanism. Decrease in the $c$-lattice parameter when lithium was added activated easier nonbasal slip dislocation of the alloys. Effects of flow stress necessary for deformation and creating a strengthening effect on the alloys were equally eliminated [51]. Plastic deformations occurring in the magnesium based alloys grains are definitely constrained by the neighboring grains and crystals which get less favorably oriented in the alloys. Ability of Mg-Li alloys and composites to deform plastically is dependent on the movement of dislocations, thereby influencing their strengthening ability and magnitude. In conclusion, density dislocation motion is among the fundamental strengthening mechanisms when applied together with other mechanisms to strengthen Mg-Li alloys and composites.

3. Conclusion
Mg-Li alloys and composites strength cannot increase infinitely. Strengthening mechanisms are applied in order to improve their properties. Each strengthening mechanism may compromise some of the alloys properties during processing. Different strengthening mechanisms and the supportive theoretical models have been outlined correlating to the microstructure and mechanical properties of the alloys. The mechanisms are highly dependent on the alloys composition and thermomechanical treatment. However no single strengthening mechanism mode absolutely strengthens the Mg-Li alloys and composites. This is because none of the mechanisms accords with the experimental results to perfection as explained by macro- and nanolevel theoretical models. Hence it is essentially important to merge different strengthening mechanisms together to design and manufacture alloys with enhanced mechanical properties.

Competing Interests
The authors declare that there is no conflict of interests regarding the publication of this paper.

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References
10 Advances in Materials Science and Engineering


