

Article

Effect of Competitive Precipitation and Texture Weakening on Mechanical Properties in a Mg-Gd-Y-Nd-Zr Alloy Processed by Integrated Multi-Directional Forging and Extrusion

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Abstract

As the lightest metallic structural material, magnesium alloys face a fundamental trade-off between strength and ductility, limiting their broader application. This study investigates a processing approach to overcome this limitation by systematically comparing the effects of direct extrusion and a multi-directional forging (MDF) combined extrusion process on a Mg-8Gd-4Y-1Nd-0.5Zr alloy. The results demonstrate that MDF pretreatment effectively refines grains and enhances dynamic precipitation. It also significantly weakens the texture, reducing the intensity from 11.14 to 3.98 and tilting the {0001} basal planes by approximately 30° from the extrusion direction. This texture weakening is attributed to the combined effects of particle-stimulated nucleation (PSN) and the orientation diversity introduced by pre-forging, which promote orientation randomization during recrystallization. The alloy processed by the combined route exhibits an excellent strength–ductility synergy in the as-extruded state, with ultimate tensile strength, tensile yield strength, and elongation reaching 315 MPa, 228 MPa, and 13.1%, respectively. After peak aging, the strength further increases to 429 MPa and 323 MPa while maintaining a ductility of 7.3%. Schmid factor analysis confirms that the combined process facilitates the activation of non-basal slip and improves strain compatibility through multi-slip activity, providing an effective pathway for developing high-performance wrought magnesium alloys.

Keywords: multi-directional forging; extrusion; texture; precipitation; mechanical properties

1. Introduction

As the lightest metallic structural material, magnesium (Mg) alloys hold great promise for applications in aerospace, automotive, rail transit, and 3C electronics owing to their low density, high specific strength, and excellent electromagnetic interference shielding effectiveness [1,2]. Nevertheless, a prominent strength ductility trade-off often exists in Mg alloys at room temperature, which severely limits their use in load-bearing components [3,4]. This limitation originates from the hexagonal close-packed (HCP) crystal structure of Mg, which offers a restricted number of active slip systems at room temperature, resulting in poor formability and pronounced mechanical anisotropy. In particular, hot plastic deforming such as extrusion and rolling frequently promote the development of a strong basal texture, which further suppresses the activation of non-basal slip systems [5,6]. Hence, achieving a simultaneous enhancement of strength and ductility in Mg alloys remains a critical challenge in the field.



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In recent years, researchers have sought to address the above-mentioned trade-off by focusing on alloy design and process control. To enhance strength, the addition of rare earth (RE) elements has proven effective in modulating microstructure and mechanical properties. Elements such as Gd and Y exhibit high solid solubility in the magnesium matrix, which decreases sharply with temperature, imparting a pronounced age-hardening response in Mg alloys [7,8]. The corresponding strengthening phases typically include GP zones, β'' , and β' precipitates [9,10]. However, alloying alone often fails to achieve a satisfactory balance with ductility. For instance, while the addition of Gd or Zn to Mg-Nd alloys can further improve strength, it frequently leads to a notable decline in plasticity [11,12]. This is primarily attributed to the formation of coarse secondary phase particles due to excessive precipitation, coupled with the enhanced basal texture that restricts the activation of non-basal slip systems. Fine-grained Mg alloys are commonly obtained via severe plastic deformation (SPD). Yang [13] optimized the extrusion process of a Mg-3.5Sm-0.6Zn-0.5Zr alloy and obtained an ultrafine-grained structure with grain sizes ranging from 470 to 500 nm. After aging, the alloy exhibited strength comparable to that of high-content Mg-RE alloys, with tensile strength, yield strength, and elongation reaching 427 MPa, 416 MPa, and 5.1%, respectively. Similarly, Wan [14] fabricated a Mg-Gd-Y-Zr alloy with a nanoscale grain structure (≈ 80 nm) using rotary forging, achieving a tensile strength of 710 MPa, a yield strength of 650 MPa, and an elongation of 4.5%. Although grain refinement can also improve ductility, the strong basal texture typically developed during hot working limits the activation of additional slip systems, thereby constraining strain compatibility and overall plasticity. Thus, it appears that gains in strength are often achieved at the expense of ductility.

Improving the plasticity of magnesium alloys generally requires the activation of non-basal slip systems or modification of the deformation texture. The addition of rare earth (RE) elements, particularly Gd and Y, has been demonstrated to effectively weaken the strong basal texture and promote the formation of a so-called “RE texture”, where the basal plane tends to tilt at a certain angle to the extrusion direction (ED) [15,16]. This texture modification facilitates basal slip during tensile loading along the ED, thereby markedly enhancing ductility. Alternatively, alloying with lightweight elements such as Li can reduce the critical resolved shear stress (CRSS) for non-basal slip, promoting the activation of prismatic $\langle a \rangle$ and pyramidal $\langle c + a \rangle$ slip systems and consequently improving plasticity [17]. Similarly, minor Si additions have been shown to activate $\langle c + a \rangle$ slip even under small strains at room temperature, leading to a significant increase in alloy ductility [18]. Beyond alloying, microstructural design offers another effective route to reconcile strength and ductility. For instance, constructing a bimodal microstructure composed of fine dynamic recrystallized (DRXed) grains and coarse unDRXed grains has emerged as a promising strategy [19]. Fine grains contribute to strength via grain boundary strengthening, while coarse grains, often accompanied by favorable textures, accommodate plastic strain. Furthermore, multi-step or hybrid deformation processing can be employed to simultaneously achieve grain refinement and texture weakening, thereby enhancing overall performance [20,21]. Multi-directional forging (MDF) effectively breaks down initial coarse grains through multi-axial strain paths, introducing high dislocation densities and dynamic precipitates that create a favorable microstructural “pre-state” for subsequent forming operations [22,23]. As a typical severe plastic deformation process, extrusion parameters directly govern the dynamic recrystallization mechanisms and texture evolution. In commercial AZ31 alloys, varying deformation conditions activate distinct slip systems and recrystallization behaviours, resulting in markedly different crystallographic orientations and mechanical properties [24,25]. These findings underscore that the inter-

play between deformation mechanisms and recrystallization behaviour lies at the core of tailoring texture and properties under complex stress states.

While the addition of Gd and Y is known to impart age-hardening effects and enable texture modification in magnesium alloys, making Mg-Gd-Y systems a focus of considerable research interest [26–28], a critical knowledge gap remains. There is still a lack of systematic understanding regarding the interplay among microstructure evolution, dynamic recrystallization mechanisms, and texture development during complex thermo-mechanical processing in these alloys, and how this interplay ultimately governs their final mechanical properties.

Therefore, this work systematically investigates an Mg-8Gd-4Y-1Nd-0.5Zr alloy subjected to MDF combined extrusion process. It aims to reveal how the forging pretreatment modulates the initial microstructure state, subsequently governs dynamic recrystallization and texture evolution during extrusion, and influences the competition between dynamic and aging precipitation. By comparing the results with those from direct extrusion, the effectiveness of this multi-stage route in achieving a superior strength–ductility balance is demonstrated. This research provides both theoretical insights into the microstructure–texture–property relationships in complex processing and a practical design strategy for developing high-performance Mg-RE alloys.

2. Materials and Methods

Mg-8Gd-4Y-1Nd-0.5Zr (wt.%) alloy ingots were fabricated via gravity casting. Prior to melting, the raw materials, including Mg-30 wt.% Gd, Mg-25 wt.% Y, Mg-30 wt.% Nd, and Mg-30 wt.% Zr master alloys, were preheated and mechanically polished to remove surface oxides. Pure magnesium was first placed in a crucible and heated to 760 °C under an argon protective atmosphere. Subsequently, the Mg-Nd, Mg-Gd, Mg-Y, Mg-Zr, and Mg-Si master alloys were added sequentially into the molten magnesium. After complete melting, the melt was stirred and treated with a refining agent for degassing and slag removal. Following settling, the melt was poured into a mold. The as-cast ingots were then subjected to homogenization treatment at 520 °C for 12, followed by water quenching.

The homogenized ingots were machined into cylindrical billets with dimensions of Φ 32 mm $\times$ 50 mm for extrusion. Extrusion was performed at a temperature of 430 °C and a speed of 1 mm/s, with an extrusion ratio of 16:1. To reduce friction, a mixture of graphite and engine oil was applied as a lubricant on the inner surface of the extrusion container. For multi-directional forging (MDF), homogenized samples were prepared into rectangular blocks measuring 40 mm $\times$ 45 mm $\times$ 50 mm. MDF was conducted on a hydraulic press at 500 °C with a strain rate of 1 s^{−1} and a true strain of 0.15 per pass. The first forging was applied on the 40 mm $\times$ 45 mm surface. Subsequently, the sample was rotated 90° and forged on the 45 mm $\times$ 50 mm surface, followed by another 90° rotation and forging on the 40 mm $\times$ 50 mm surface, completing one full deformation cycle. After 9 passes forging, the sample was immediately water-quenched and subsequently used as the extrusion billet under the same extrusion parameters described above. The schematic diagram of MDF and extrusion processes is shown in Figure 1.

Tensile specimens with a gauge diameter of 5 mm, gauge length of 25 mm, and total length of 75 mm were machined for mechanical testing. Room-temperature tensile tests were performed using an Instron 3369 testing machine (Instron Corporation, Boston, MA, USA). Each set of experiments included three parallel samples. For hardness measurements, samples were sequentially ground with 600#, 1200#, and 1500# grit SiC paper, dried, and tested on an HV-10 Vickers hardness tester (Aolong, Shanghai, China) under a load of 3 kg with a dwell time of 30 s. The reported hardness values represent the average of five independent measurements.

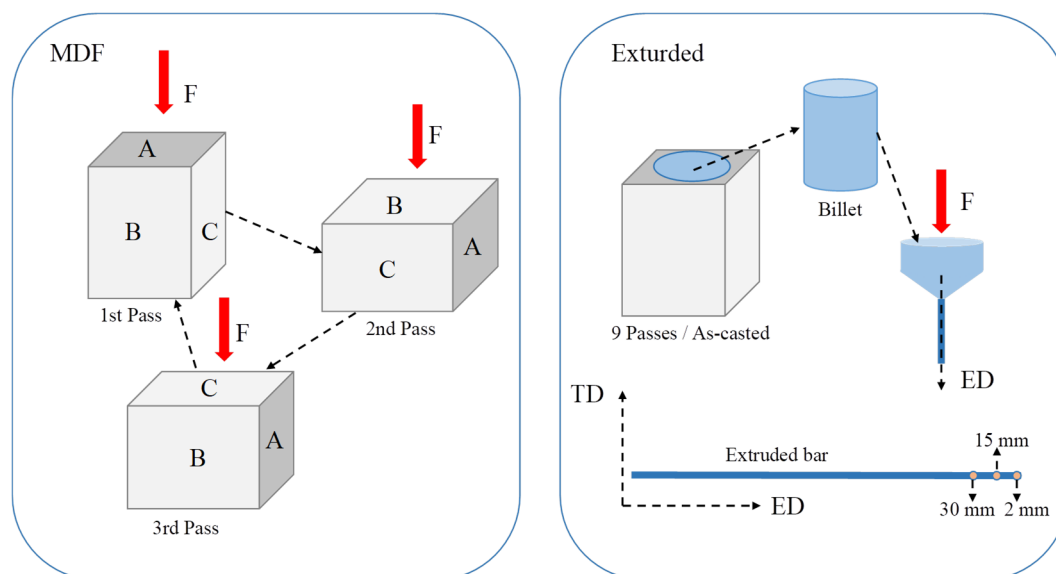


Figure 1. Schematic diagram of MDF and extrusion processes.

Metallographic samples were mechanically ground, polished, and etched for 10~15 s using a solution composed of 4 mL nitric acid, 5 g citric acid, and 96 mL ethanol. For scanning electron microscopy (SEM), samples were prepared following the same grinding and polishing procedure. Microstructural and secondary phase characterization was performed using a FEI Quanta-200 SEM (FEI, Hillsboro, OR, USA) equipped with an energy dispersive spectroscopy (EDS) system. Electron backscatter diffraction (EBSD) samples were prepared similarly to metallographic specimens. EBSD analysis was conducted using an Oxford detector mounted on a HELIOS NanoLab 600i scanning electron microscope (FEI, Hillsboro, OR, USA), and data were processed with HKL Channel 5 software (V5.12). Transmission electron microscopy (TEM) samples were first mechanically ground to approximately 200 μm , followed by manual thinning to about 70 μm . Discs of Φ 3 mm were punched and then electropolished using a twin-jet electrolytic polisher (MTP-1) with a solution containing 2.81 g LiCl, 5.91 g $\text{Mg}(\text{ClO}_4)_2$, 265 mL methanol, and 53 mL butyl cellosolve, cooled to -40°C with liquid nitrogen under a current of 15 mA. TEM observations were conducted on a Tecnai G2 20 microscope (Thermo Fisher Scientific, Hillsboro, OR, USA).

3. Results

Figure 2 presents the metallographic structures of the Mg-8Gd-4Y-1Nd-0.5Zr alloy. In the as-cast condition, the microstructure is characterized by equiaxed dendrites with an average grain size of approximately 40 μm . Additionally, a significant quantity of intergranular non-equilibrium eutectic phases, exhibiting a network-like morphology, is observed along the grain boundaries. Following solid solution treatment, the majority of these grain boundary eutectic phases dissolve, resulting in their virtual elimination. However, a limited number of blocky second-phase particles persist within the grain interiors. Moreover, a notable grain growth occurs during the treatment, increasing the average grain size to approximately 100 μm .

The SEM images and corresponding EDS results of the Mg-8Gd-4Y-1Nd-0.5Zr alloy in both the as-cast and solid-solution states are shown in Figure 3. As shown in Figure 3a, the non-equilibrium eutectic phase in the as-cast microstructure exhibits a distinctive “skeletal” morphology at higher magnification (highlighted by the red box). EDS analysis indicates that these phases are primarily enriched with rare earth (RE) elements. Following a solid-solution treatment, the continuous network-like eutectic structure was almost entirely dissolved into the matrix. However, a small number of blocky secondary phases remained

undissolved shown in Figure 3b. EDS analysis confirmed that these residual phases were also RE-rich. The measured atomic ratio of Mg to (Gd and Y) in these particles was approximately 5:1. Given the similar chemical behavior of Gd and Y in magnesium alloys, where these elements can readily substitute for each other in the secondary phase [29], these phases are identified as $\text{Mg}_5(\text{Gd}, \text{Y})$.

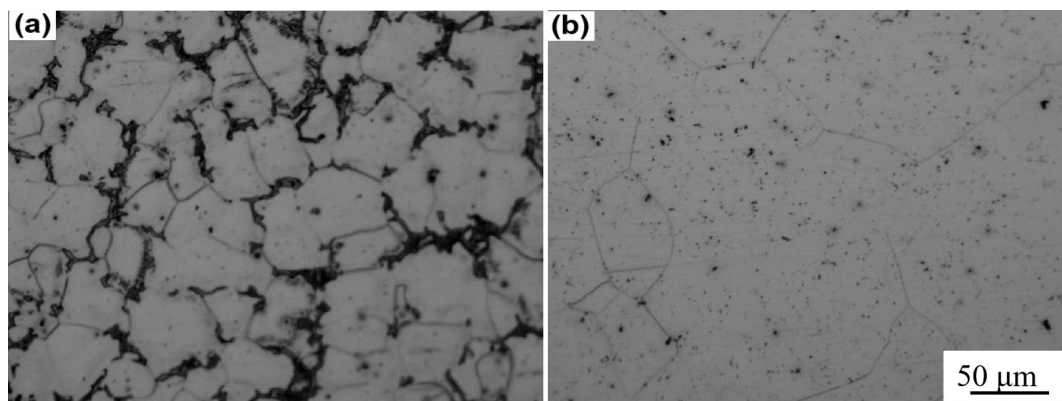


Figure 2. Metallographic structures of the Mg-8Gd-4Y-1Nd-0.5Zr alloy. (a) As-cast condition; (b) solid solution-treated condition.

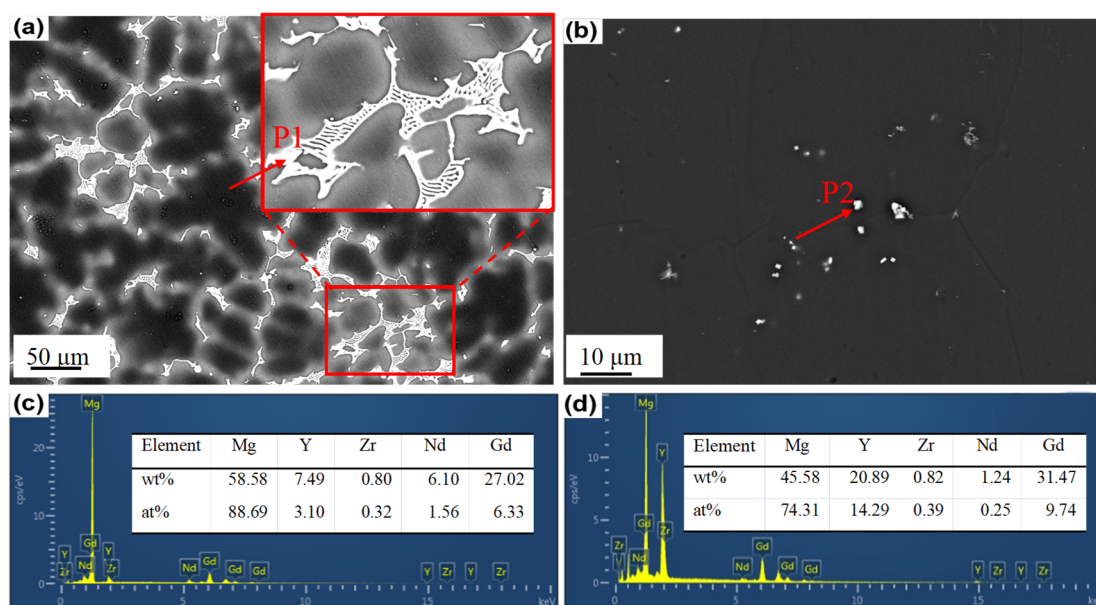


Figure 3. SEM and EDS analysis of Mg-8Gd-4Y-1Nd-0.5Zr alloy. (a) As-cast state; (b) solid solution state; (c) EDS analysis of P1; (d) EDS analysis of P2.

Figure 4a is the EBSD analysis of the Mg-8Gd-4Y-1Nd-0.5Zr alloy after multi-directional forging. As shown, the grain boundaries exhibit a distinct “serrated” morphology following 9 forging passes. This characteristic is attributed to strain induced grain boundary migration, wherein a difference in stored deformation energy between adjacent grains causes originally straight boundaries to bulge toward the grain with higher energy storage, thereby creating a serrated profile and providing favorable conditions for recrystallization. Additionally, fine equiaxed recrystallized grains are observed to be distributed in a chain-like manner along the boundaries of coarse grains. This partially recrystallized microstructure is expected to significantly influence the microstructure and mechanical properties during subsequent extrusion. Combined with SEM observation presented in Figure 4b, dynamic precipitation is also evident in the alloy during forging. This

finding appears inconsistent with the report by Xiao [30], which indicated that no dynamic precipitation occurs when compression temperature exceeds 450 °C, as alloying elements such as Gd and Y remain in solid solution. Although the initial forging temperature in this study was 500 °C, the temperature of the magnesium billet continuously decreased during multi-pass forging. By the completion of the 9th pass, the billet temperature had dropped to approximately 420 °C. At this lower temperature, Gd and Y atoms exhibit limited solubility in the matrix and can diffuse to form dynamic precipitates.

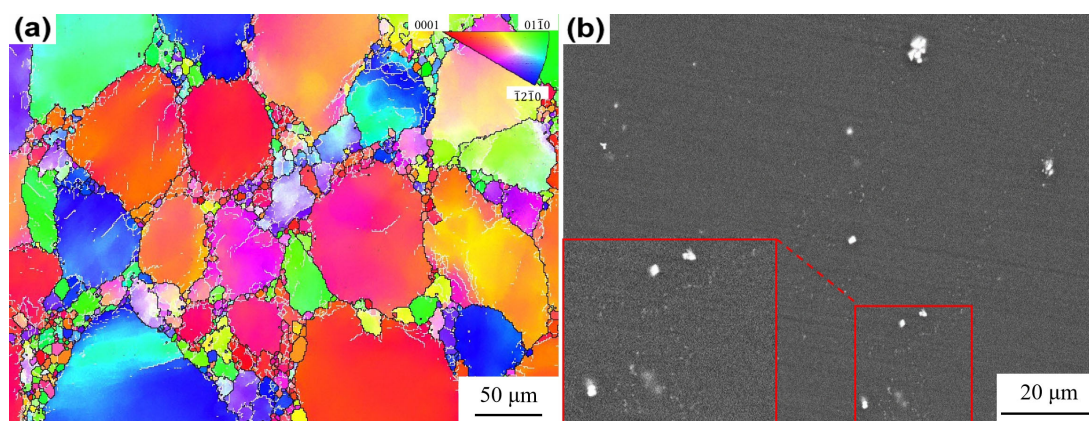


Figure 4. Microstructure of Mg-8Gd-4Y-1Nd-0.5Zr alloy after forging. (a) EBSD and (b) SEM.

Figure 5 presents the SEM microstructures of the alloy after extrusion alone and MDF followed by extrusion. In both processing conditions, the alloy undergoes extensive plastic deformation, accompanied by significant dynamic precipitation of secondary phases. However, the alloy subjected to MDF prior to extrusion exhibits a considerably higher density of fine precipitates compared to that processed by direct extrusion. With higher imposed strain than forging, extrusion generates a substantially greater dislocation density within the alloy. The combination of this high dislocation density and the elevated extrusion temperature provides both diffusion pathways (dislocation pipes) and the necessary driving force for solute atom diffusion. Consequently, the extrusion deformation promotes more extensive dynamic precipitation than the forging process alone.

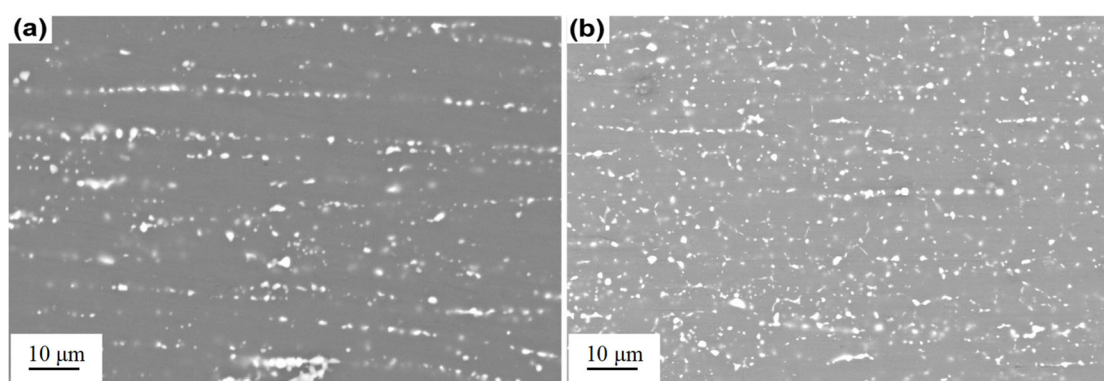


Figure 5. SEM results of Mg-8Gd-4Y-1Nd-0.5Zr alloy after hot deformation. (a) Extrusion; (b) MDF + extrusion.

In the case of the MDF + extrusion route, the initial MDF step itself introduces a high dislocation density and fine-scale dynamic precipitates into the alloy. This pre-deformed and pre-precipitated microstructure plays a crucial role during the subsequent extrusion. Firstly, the pre-existing high dislocation density provides abundant nucleation sites for new precipitates, significantly enhancing the nucleation rate. Secondly, the fine precipitates from

MDF pin dislocations, helping to retain a high dislocation density during extrusion, while their partial dissolution also supplies solute atoms for continuous precipitation. Finally, the well-developed dislocation network significantly accelerates solute diffusion through pipe diffusion mechanisms. Collectively, this “pre-conditioned” microstructure resulting from MDF enables earlier initiation, faster kinetics, and more uniform distribution of dynamic precipitation during the subsequent extrusion. Therefore, the MDF + extrusion processing path ultimately produces a finer and more dispersed precipitate distribution compared to direct extrusion alone.

The EBSD analysis of the alloys processed by direct extrusion and MDF + extrusion is presented in Figure 6. The grains subjected to MDF pretreatment are significantly refined. Statistical analysis of grain size reveals that the directly extruded alloy exhibits a maximum grain size of 28 μm and an average grain size of 8.7 μm . In contrast, the alloy processed by MDF prior to extrusion shows a much finer microstructure, with a maximum grain size of 16 μm and an average grain size of 4.8 μm . Notably, the different thermo-mechanical routes also lead to distinct texture characteristics. As shown in Figure 6e, the direct extruded alloy displays a texture with $\{0001\}$ planes oriented perpendicular to the extrusion direction and a maximum texture intensity of 11.14. In comparison, the MDF + extrusion processed alloy exhibits a weakened texture, with the $\{0001\}$ basal planes tilted approximately 30° away from the extrusion direction and a significantly reduced intensity of 3.98 shown in Figure 6f.

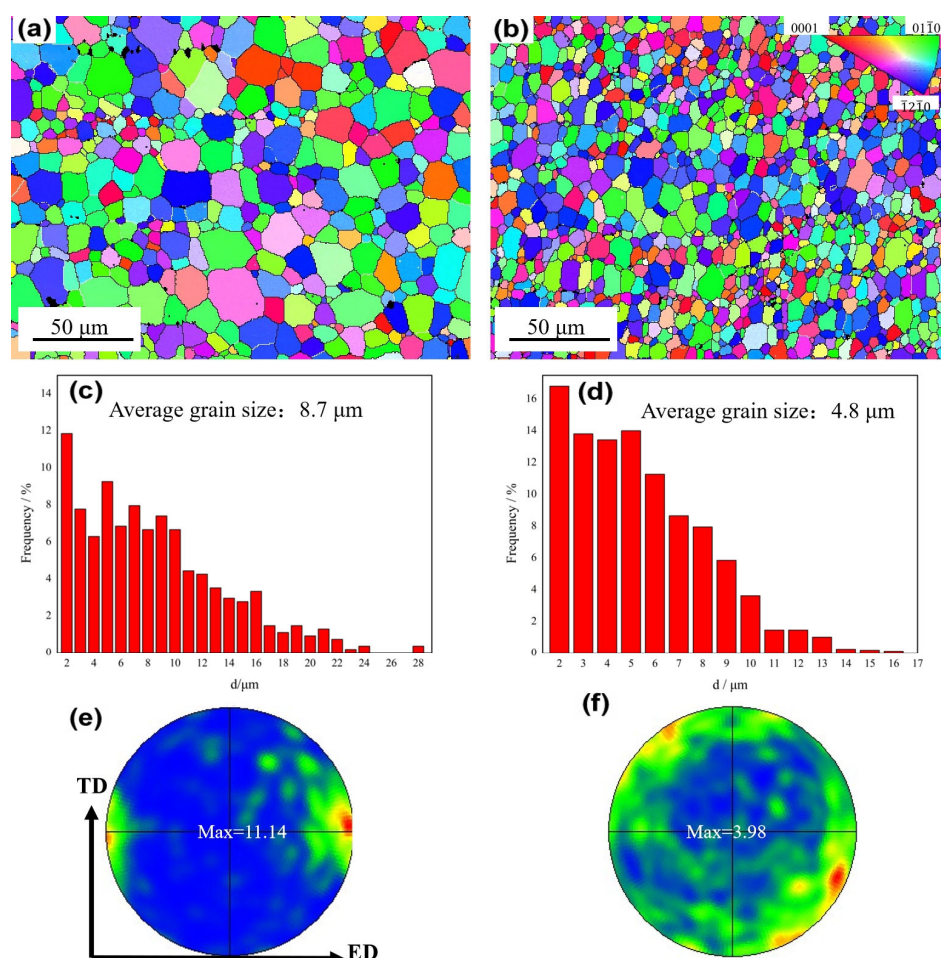


Figure 6. EBSD results of Mg-8Gd-4Y-1Nd-0.5Zr alloy after hot deformation. (a,c,e) Extrusion; (b,d,f) MDF + extrusion.

Figure 7 presents the age-hardening curves of the directly extruded and MDF + extrusion alloys at 215 °C. The hardness increases rapidly during the initial aging stage and reaches its peak value at approximately 22 h for both alloys. Further extension of the aging time leads to a decrease in hardness, indicating an over aging. Interestingly, in the as-deformed state prior to aging, the directly extruded alloy exhibits a lower hardness of 79.9 ± 0.99 HV, while the alloy subjected to MDF + extrusion shows a higher initial hardness of 88.65 ± 2.32 HV. During the aging process, the hardness of the directly extruded alloy increases more significantly, reaching a peak hardness of 143.16 ± 0.78 HV at peak aging, which is 3.21 HV higher than that of MDF + extrusion alloy.

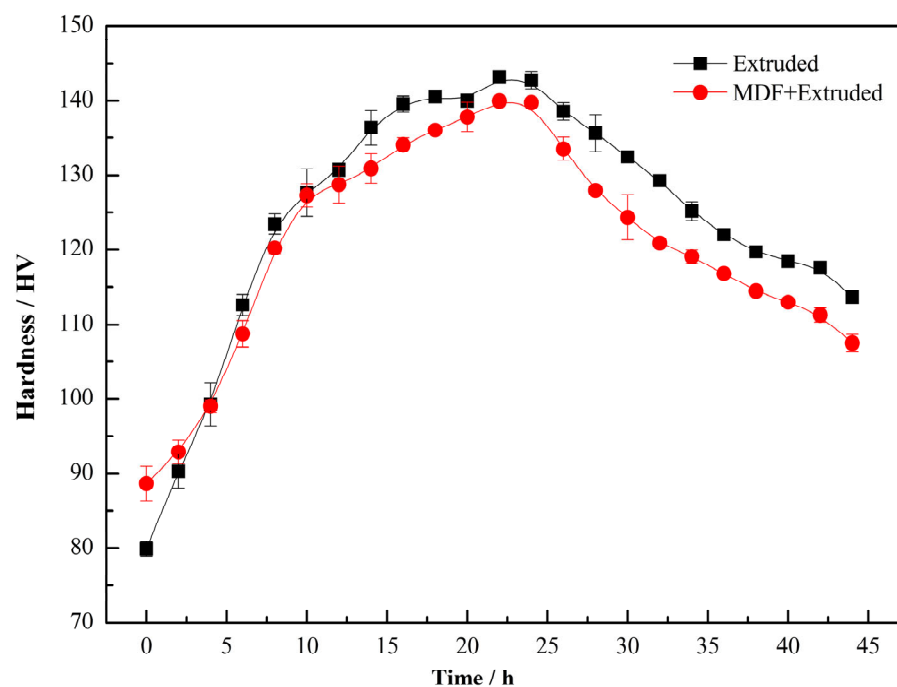


Figure 7. Age-hardening curves.

Figure 8 presents the room temperature tensile properties of the alloys processed by extrusion and MDF + extrusion at peak aged state. The tensile strength of MDF + extrusion alloy is 315 ± 2.1 MPa, the yield strength is 228 ± 3.6 MPa, and the elongation is $13.1 \pm 1.7\%$ in the as-deformed state. After peak aging treatment, its tensile strength and yield strength increases to 429 ± 3.2 MPa and 323 ± 1.8 MPa, while the elongation decreases to $7.3 \pm 1.3\%$. In comparison, the directly extruded alloy exhibits tensile strength, yield strength and elongation values of 293 ± 2.1 MPa, 211 ± 1.9 MPa, and $8.3 \pm 2.2\%$, respectively, in the unaged condition. Following aging, these values change to 433 ± 2.4 MPa, 332 ± 2.5 MPa, and $4.43 \pm 0.9\%$. Notably, the MDF + extrusion alloy demonstrates superior ductility in both the as-deformed and peak-aged states compared to its directly extruded counterpart. A further comparison reveals that the MDF + extrusion alloy possesses higher strength in the as-deformed condition. However, after aging, its strength becomes slightly lower than that of the directly extruded alloy, which is consistent with the age-hardening curves presented earlier.

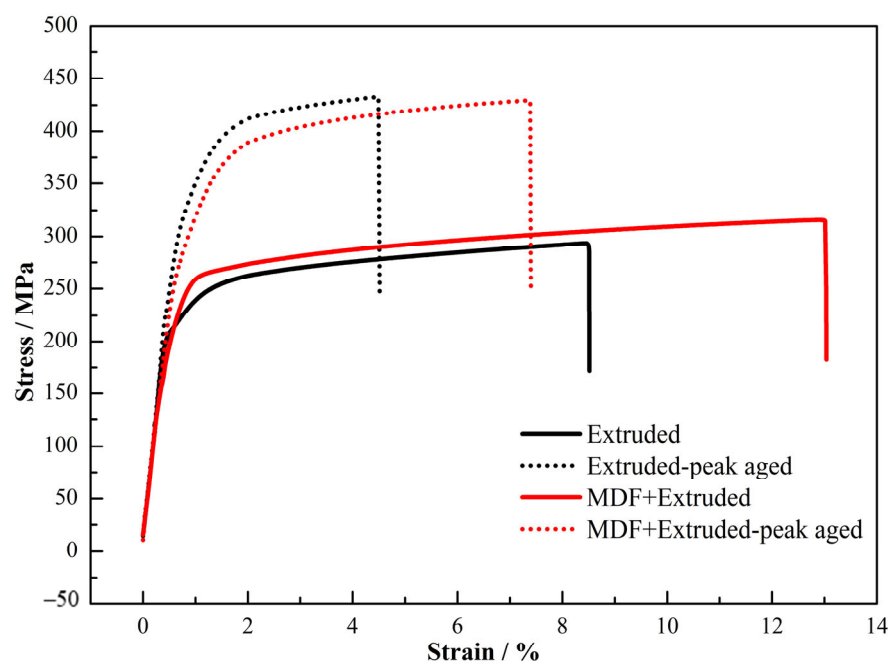


Figure 8. Results of room temperature tensile test.

As shown in Figure 9, TEM observations reveal a high density of nanometer scale β' precipitates in both alloys under the peak aged condition. The β' phase possesses a base centered orthorhombic structure [31], with lattice parameters of $a = 2a_{\text{Mg}} = 0.64$ nm, $b = 8d_{\{10\bar{1}0\}_{\text{Mg}}} = 2.22$ nm, $c = c_{\text{Mg}} = 0.52$ nm. Its orientation relationship with the magnesium matrix is defined as $[001]_{\beta'} \parallel [0001]_{\text{Mg}}$, $\{100\}_{\beta'} \parallel \{2\bar{1}\bar{1}0\}_{\text{Mg}}$. A comparison of the precipitate characteristics between the directly extruded and the MDF + extruded alloys after peak aging treatment shows no distinct difference in morphology. Both exhibit a characteristic “dumbbell-shaped” composite structure, which consists of a relatively thick ellipsoidal core flanked symmetrically by slender and needle-like extensions. The central ellipsoidal region maintains a coherent or semi-coherent interface with the Mg matrix, while the thin needle-like ends grow along low strain energy crystallographic directions to release long-range elastic strain [32]. However, differences are observed in terms of precipitate size and number density. In the extruded alloy, the β' precipitates have an approximate length of 20 nm and width of 5 nm. In contrast, those in the MDF + extruded alloy are coarser, with an average length of about 28 nm and width of about 10 nm. Furthermore, the number density of the β' precipitates is noticeably lower in the MDF + extruded alloy compared to the extruded counterpart. This difference is attributed to the extensive dynamic precipitation occurring during the multi-directional forging process, which consumes a significant portion of the rare earth solute atoms in the matrix. Consequently, the solid solution supersaturation prior to aging is reduced in the MDF + extruded alloy. During the subsequent peak aging treatment, the lower reserve of available solute atoms leads to a reduced nucleation rate for precipitation hardening phases, resulting in fewer precipitates. This lower nucleation rate also facilitates the migration and aggregation of solute atoms toward fewer precipitate nuclei, thereby promoting precipitate coarsening.

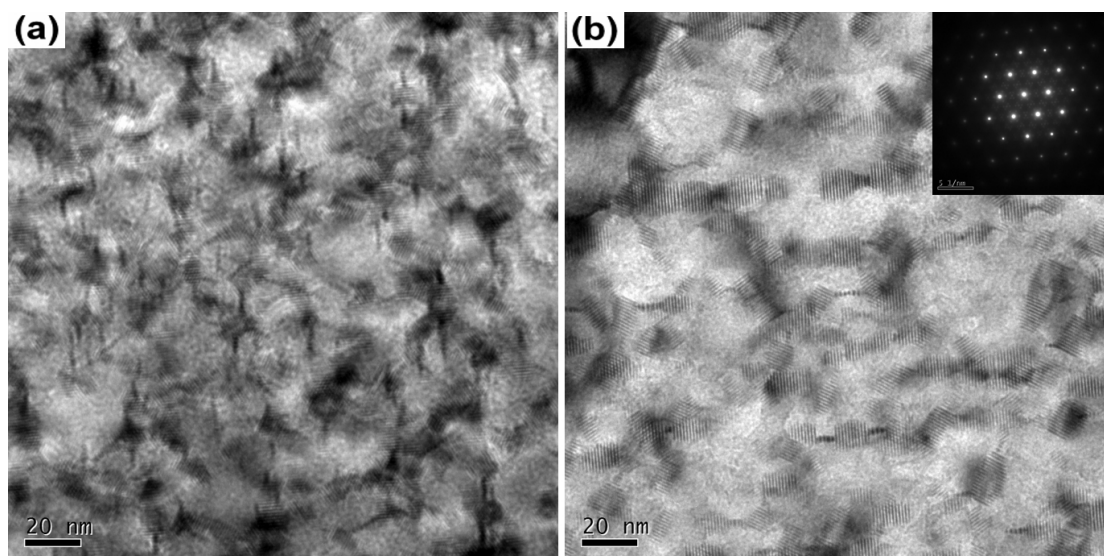


Figure 9. TEM images of peak aged state. (a) Extrusion; (b) MDF + extrusion.

4. Discussion

4.1. Strength Analysis

Under the two processing routes mentioned above, the alloys exhibit distinct strength characteristics after peak aging treatment: although the MDF + extrusion process endows the alloy with higher strength in the as-deformed state, the strength after aging is slightly lower than that of the directly extruded counterpart. This seemingly contradictory phenomenon can be reasonably explained by the competition between dynamic precipitation and aging precipitation.

During the deformation process, dynamic precipitation significantly promotes due to more complex strain path and higher accumulated strain in the MDF + extrusion alloy. A large number of fine dynamic precipitates form in situ during deformation, as shown in Figure 5. In room temperature tensile tests, these finely dispersed secondary phases contribute to strength by effectively impeding dislocation motion. The strengthening contribution of these micron-scale dynamic precipitates can be assessed via the classical Orowan bypassing mechanism [33]. It is quantified using a recently proposed expression, given here as Formula (1) [34]:

$$\Delta\sigma_d = \left(M \frac{0.2Gb}{\pi\sqrt{(1-\nu)}} \right) \frac{\ln(2\sqrt{2/3}r/b)}{\lambda} \quad (1)$$

Here, M , G , b , and ν represent the Taylor factor, the shear modulus, the Burgers vector, and Poisson's ratio, all of which are intrinsic material properties. The parameters r and λ denote the average particle radius and interparticle spacing, respectively. According to the Orowan equation, the strengthening contribution increases with larger particle size r and smaller interparticle spacing λ . In the present study, while the size of dynamic precipitates shows no significant difference between the extruded and MDF + extruded alloys, the MDF + extruded material exhibits a notably higher number density of precipitates. This results in a considerably reduced interparticle spacing, thereby enhancing the Orowan strengthening contribution. Furthermore, the MDF + extruded alloy possesses finer grains, leading to a more pronounced grain boundary strengthening effect. Consequently, under the combined effects of precipitation strengthening and grain refinement, the MDF + extruded alloy in the as-deformed state demonstrates higher strength than its directly extruded counterpart.

After aging treatment, β' precipitates formed in the alloys processed via both hot plastic routes. Research by Bhattacharyya [35] demonstrated that prismatic β' precipitates can be sheared by basal $\langle a \rangle$ dislocations. Therefore, the strengthening contribution from these fine precipitates can be characterized by a shearing mechanism [36]:

$$\tau_{cut} = \alpha f^{1/2} r^{1/2} \quad (2)$$

where α is a constant, f the volume fraction, and r the average radius of the β' precipitates. According to the TEM observations shown in Figure 8, after peak-aging treatment, the β' precipitates in the directly extruded alloy are finer and possess a higher number density, leading to a more significant age-hardening effect. Compared to the MDF + extruded alloy, the directly extruded alloy contains fewer dynamic precipitates, thereby retaining a higher supersaturation of solute atoms which facilitates the formation of a dense distribution of fine β' precipitates during the subsequent peak-aging treatment. Furthermore, the superior strength achieved in the extruded alloy after aging indicates that precipitation strengthening plays the predominant role in enhancing the overall strength, surpassing the contributions from both dispersion strengthening and grain boundary strengthening.

4.2. Formation of Weak Texture

The texture of the MDF + extrusion transformed from the strong texture of direct extrusion to a weakened texture with the basal plane tilted approximately 30° from the extrusion direction. To better understand the origin of this texture evolution in the MDF + extrusion alloy, it is essential to analyze the microstructural changes and texture development during the extrusion process, particularly in its initial stages. EBSD analysis was performed on samples taken from three locations along the extrusion direction at distances of 2 mm, 15 mm, and 30 mm (the locations were shown in Figure 1) from the extrusion start point, as presented in Figure 10.

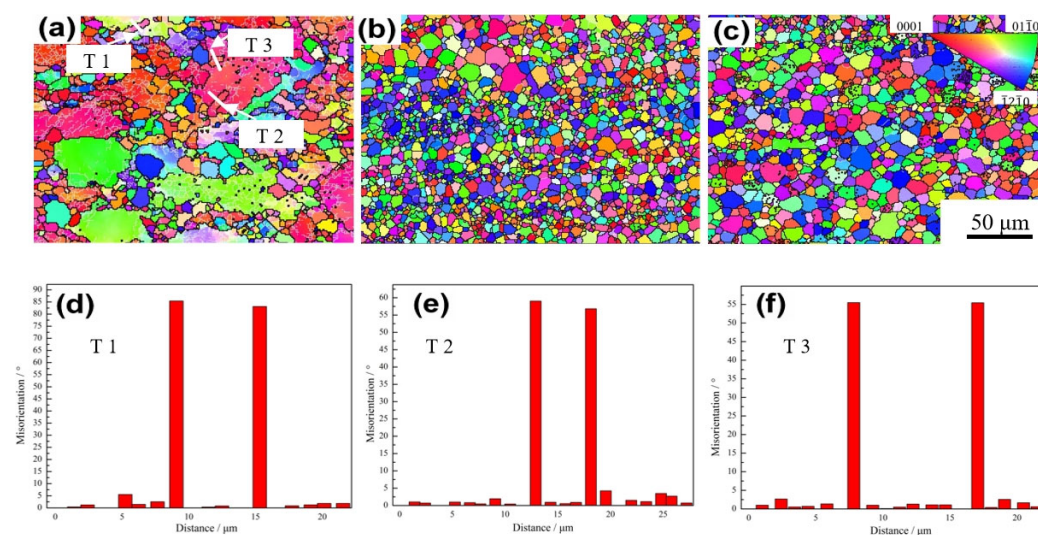


Figure 10. EBSD observations from the initial MDF + extrusion stage. (a) 2 mm; (b) 15 mm; (c) 30 mm; (d–f) misorientation of T1, T2 and T3 shown in (a).

At the initial stage (2 mm), the alloy exhibits a low degree of recrystallization. Most of the original grains become twisted and elongated, changing from an elliptical or circular morphology in the forged state to a bent and slender shape. Additionally, twinning is observed within the grains, as indicated by the white arrows in Figure 10a. Orientation analysis of these twins reveals that Twin 1, with a misorientation of approximately 86° relative to the matrix, is identified as a tensile twin, while Twins 2 and 3, with misorientations

of 64° and 56° respectively, are identified as compression twins [37–41]. During round bar extrusion, the material is subjected to triaxial compressive stress, which facilitates the formation of compression twins. The critical resolved shear stress (CRSS) required to activate compression twins ranges from 76 to 153 MPa, significantly higher than the 2–2.8 MPa required for tensile twins [42]. The formation of compression twins in Figure 10a indicating that the material experiences substantial loading at this stage. Moreover, due to the relatively large grain size and extended slip distance, twinning becomes necessary to accommodate the imposed deformation. Although the shear strain introduced by twinning is limited and generally contributes little to the plasticity of magnesium alloys, it can effectively alter the grain orientation. This reorientation changes the Schmid factor (SF) of the grains, thereby promoting further slip activity. Additionally, during deformation, dislocations tend to pile up at compression twin boundaries, creating stress concentrations that can stimulate recrystallization. In contrast, tensile twins primarily undergo grain boundary migration during deformation [43–45]. As deformation progresses to 15 mm and 30 mm, the microstructure exhibits fully recrystallized characteristics. However, the grains at the 30 mm location show noticeable growth compared to those at 15 mm.

Integrating the distinct microstructural characteristics before and after the two deformation processing routes, an analysis of recrystallization and texture evolution during the MDF + extrusion hybrid forming process was conducted and shown in Figure 11, aiming to explore the underlying mechanism for its weakened texture formation. The material initially undergoes MDF can induces partial dynamic recrystallization, resulting in a refined grain structure and generating numerous dynamic precipitates within the matrix. These precipitates subsequently act as particle-stimulated nucleation (PSN) sites during the following extrusion, promoting the nucleation of recrystallized grains with non-basal orientations.

At the initial extrusion stage of 2 mm, the imposed strain is relatively low, leading to a limited extent of recrystallization. Although the overall recrystallization fraction remains small due to insufficient accumulated strain, the PSN mechanism is already active, with newly nucleated recrystallized grains forming at precipitate interfaces and exhibiting random orientations. However, the overall texture is still dominated by the unrecrystallized grains, which become twisted under the high extrusion force, leading to a regionally consistent grain orientation and thus a strong texture. As extrusion proceeds to the 15 mm position, the combined effects of increased strain and thermal activation become significant, and dynamic recrystallization progresses extensively. The concurrent operation of PSN, continuous dynamic recrystallization (CDRX), and discontinuous dynamic recrystallization (DDRX) results in the formation of grains with increasingly random orientations. This effectively dilutes and weakens the originally strong texture, evolving the overall texture into one with lower intensity and a more scattered distribution. Notably, since the recrystallized microstructure becomes dominant at this stage, the texture of all grains shows a high degree of consistency with that of the recrystallized grains, both exhibiting a weak and scattered orientation distribution. At the 30 mm location, the texture morphology remains fundamentally similar to that at 15 mm, retaining its weak and scattered characteristics. However, a noticeable grain coarsening is observed, indicating that the extrusion process has entered a stable stage.

Nevertheless, the texture distribution in the MDF + extruded material shown in Figure 6 is not entirely random but shows a certain degree of orientation clustering. The formation of this clustering may be related to the grain growth process following recrystallization completion. During grain boundary migration, orientations with energetic advantages or better strain compatibility may be selectively retained and grow, forming orientation clusters within the overall weakened texture background [46]. Additionally, the

memory effect of the initial texture and the differential activation of slip systems during deformation could also contribute to a certain structural correlation among the recrystallized grain orientations [47].

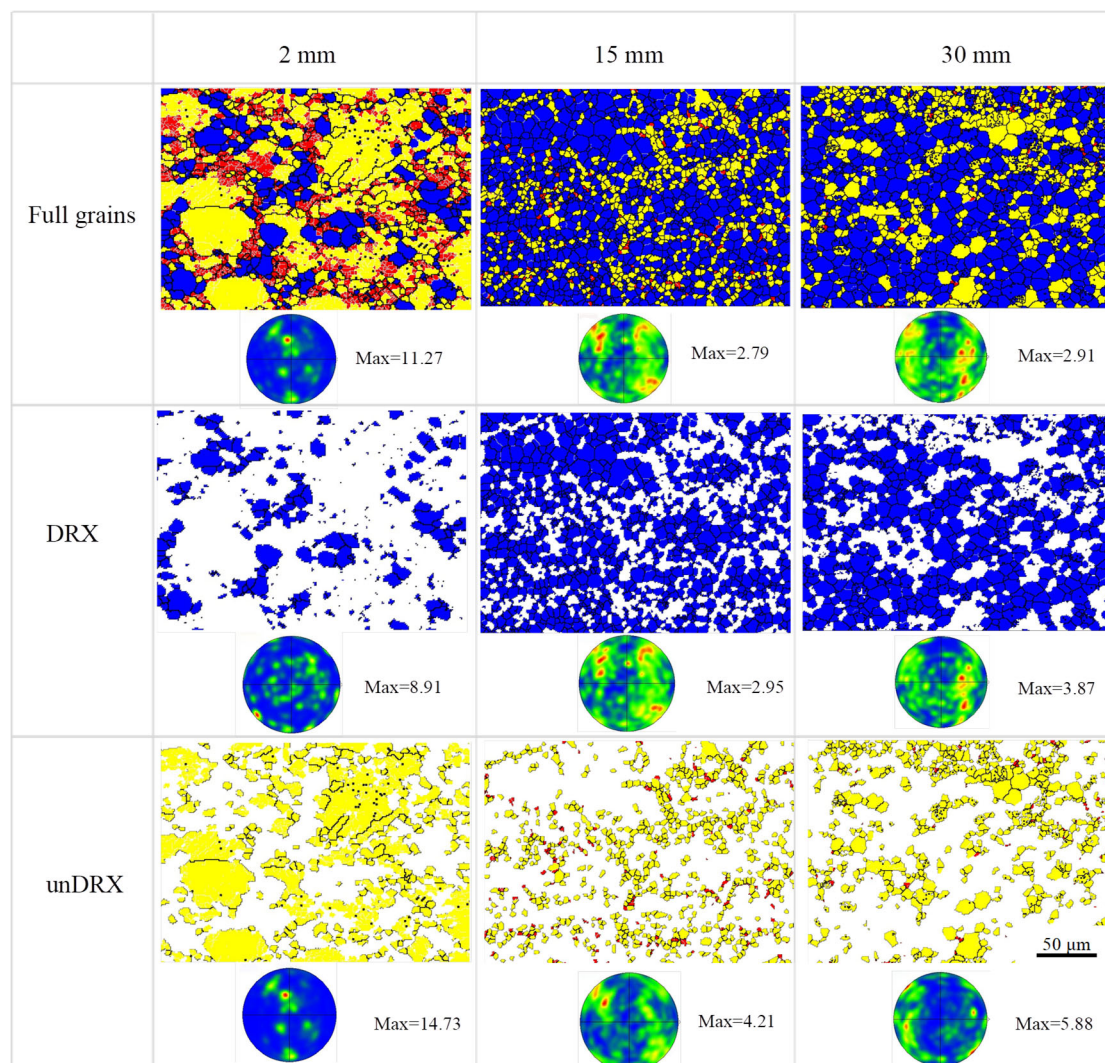


Figure 11. Microstructure and texture evolution of the initial MDF + extrusion stage.

In contrast, in the direct extrusion process, due to the lack of pre-deformation and sufficient precipitates, dynamic recrystallization is primarily dominated by grain boundary bulging. Consequently, the newly formed grains largely inherit the strong basal orientation of the original coarse grains. Under a uniaxial strain path, grains gradually rotate to a specific preferred orientation, resulting in a texture with high intensity and concentrated distribution. In comparison, the MDF + extrusion process provides a solid foundation for weakened texture formation through its abundant dynamic precipitates, pre-deformation induced fine grained structure, and the associated random orientations.

4.3. Ductility Analysis

The room temperature tensile properties indicate that the alloy processed by MDF + extrusion exhibits a significantly higher elongation than its directly extruded counterpart in both the as-extruded and peak aged conditions. The ductility of magnesium alloys is critically dependent on the activation of slip systems, which is closely related to the SF. Figure 12 presents the SF distributions for various slip systems in the directly extruded and MDF + extrusion processed alloys.

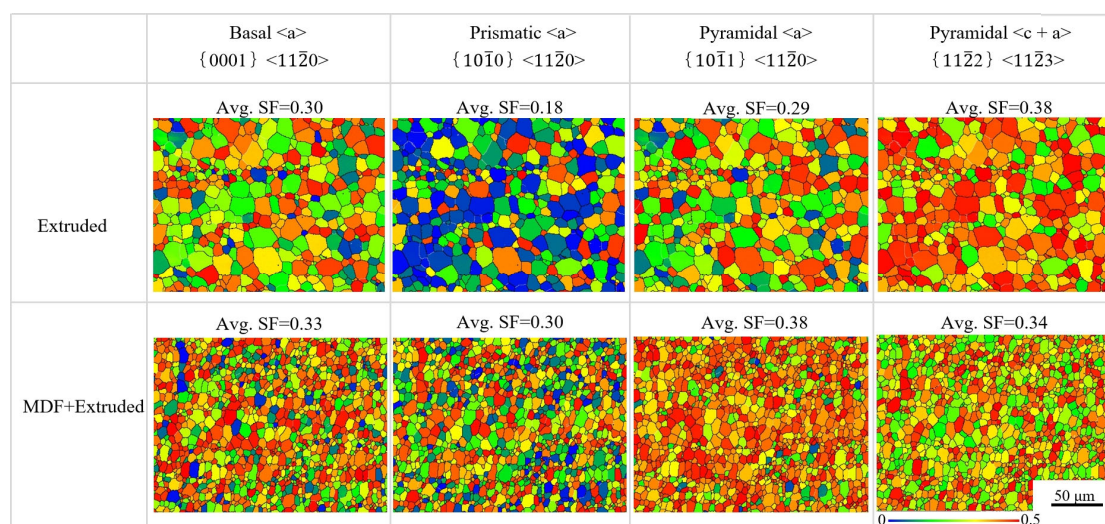


Figure 12. Schmidt factors of slip systems in extruded and MDF + extruded alloys.

For the directly extruded alloy, the average SFs for basal $\langle a \rangle$, prismatic $\langle a \rangle$, pyramidal $\langle a \rangle$, and pyramidal $\langle c + a \rangle$ slip systems are 0.30, 0.18, 0.29, and 0.38, respectively. In contrast, the corresponding average values for the MDF + extrusion alloy are 0.33, 0.30, 0.38, and 0.34. The comparison reveals that while the average Schmid factor for the pyramidal $\langle c + a \rangle$ slip decreases by 0.04 in the MDF + extrusion alloy, all $\langle a \rangle$ type slip systems show a notable increase. Specifically, the pyramidal $\langle a \rangle$ slip increases by 0.09, the prismatic $\langle a \rangle$ slip increases by 0.12, and the basal $\langle a \rangle$ slip increases by 0.03. Importantly, in the MDF + extrusion alloy, the average SFs for all slip systems exceed 0.30, suggesting that all these systems can be readily activated during tensile deformation [48]. Conversely, in the directly extruded alloy, only the basal $\langle a \rangle$ and pyramidal $\langle c + a \rangle$ slip systems have average SFs above 0.30. Consequently, the MDF + extrusion processed alloy can activate a greater variety of slip systems under tensile loading, enabling more homogeneous strain accommodation and thereby contributing to its superior ductility.

5. Conclusions

1. The MDF pretreatment induced dynamic precipitation of the Mg-8Gd-4Y-1Nd-0.5Zr alloy. This provided favorable conditions for PSN during the subsequent extrusion process, promoting recrystallization nucleation and further grain refinement. Consequently, the average grain size was reduced from 8.7 μm in the directly extruded condition to 4.8 μm after MDF + extrusion.
2. The MDF + extrusion process resulted in a significantly weakened texture with a maximum intensity of 3.98, compared to 11.14 in the directly extruded material. Additionally, the basal plane was tilted approximately 30° away from the extrusion direction. This texture modification facilitated the activation of non-basal slip systems during deformation.
3. A competitive relationship exists between dynamic precipitation and aging precipitation. The extensive dynamic precipitation during MDF consumed a significant amount of solute atoms in the matrix, which led to coarsening and a reduced number density of β' precipitates during subsequent peak aging. As a result, the age-hardening effect was slightly lower in the MDF + extruded alloy compared to the directly extruded one. The MDF + extruded alloy exhibited a tensile strength of 315 MPa and a yield strength of 228 MPa, attributed to the combined strengthening effects of grain refinement and dispersion strengthening from dynamic precipitates.

4. The weakened texture significantly enhanced the strain accommodation capability of the alloy. In the MDF + extruded alloy, the average SFs for all major slip systems exceeded 0.3, promoting the activation of multiple slip systems during tensile deformation. This enabled more homogeneous strain distribution, resulting in an elongation of 13.1% in the as-deformed state and 7.3% in the peak-aged condition, both of which are substantially higher than those achieved by direct extrusion.

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