

Optimization of microstructure and properties in an as-extruded Mg-2Zn-2Al-0.3Mn solid-solution alloy via yttrium addition

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Abstract

The development of high-performance magnesium (Mg) alloys requires precise microstructural control through composition design and thermomechanical processing. While yttrium (Y) addition is known to benefit Mg alloys, its role in dynamic recrystallisation (DRX) mechanisms and the resulted synergy of mechanical and corrosion properties within the MgZnAlMn system remains elusive, especially when integrated with solution treatment and extrusion. In this work a systematic investigation is performed to study the effect of Y content (0–5.0 wt.%) on the solution-treated and extruded Mg_{2.0}Zn_{2.0}Al_{0.3}Mn alloy. The dominant process of DRX from twin-induced DRX (TDRX) at 0 % Y gradually transits to discontinuous and continuous DRX (DDR_X, CDR_X), particle-stimulated nucleation (PSN), and the Smith-Zener pinning effect in Y-containing alloys. This transition is driven by the formation of Al₂Y and (Al, Zn)₁₁Y₃ precipitates and leads to continuous grain refinement and twin suppression. An optimized Y content is identified at 3.0 wt.% to reach a strength- plasticity ultimate tensile strength of 289.4 MPa, yield strength of 214.1 MPa and elongation of 25.8 % along with the best corrosion resistance.

Key words: magnesium alloy, microstructures, mechanical properties, corrosion resistance

1. Introduction

Accelerated global industrialization and energy crisis demands lightweighting of structural materials to benefit energy conservation, emission reduction, and sustainability [1]. In this regard, Mg alloys offer potential solutions to weight reduction in the automotive, aerospace, and computer, communication, and consumer electronics industries [2, 3]. However, conventional Mg alloys suffer from insufficient ambient-temperature mechanical properties and poor corrosion resistance, yet severely hindering their widespread application as structural materials [4, 5]. Thus, new strategies to improve mechanical properties and corrosion resistance of Mg alloys are demanded.

In recent years, MgZnAlMn system alloys have garnered significant research interest due to their low

cost and favorable processability [6, 7]. However, the second phase particles constituted coarse β -Mg₁₇Al₁₂ tend to form during solidification in these alloys. Their appearance deteriorates both mechanical properties and corrosion resistance [8, 9]. To enhance mechanical and anticorrosive properties of MgZnAlMn alloys, Y has been widely added due to its unique electron layer structure and high solid solubility in the Mg matrix. Notably, Y can interact with Al and Zn to form thermostable MgZnY quasicrystalline phases or Al₂Y intermetallic compounds. Meanwhile, the precipitation of the low-melting-point β -Mg₁₇Al₁₂ phase is simultaneously suppressed. The synergies between Y-compound formation and suppression enhance the alloy's performances [10, 11]. Nevertheless, microstructural defects such as inhomogeneity, porosity, cracking, and segregation widely exist in cast Mg

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alloys and limit their practical engineering applications [12]. Applying Y-addition to cast Mg alloys may lead to betterment of the alloy performance.

Currently, comprehensive property enhancement of cast Mg alloys primarily involves solid solution treatment and extrusion deformation processing [13]. Yuan et al. demonstrated improved mechanical properties of solid solution treated Mg1Al12Y alloy at 540 °C for 4 h thanks to partial dissolution of the Mg₂₄Y₅ phase, promoted solute atom diffusion, and precipitation of 18R-LPSO phases [14]. Li et al. investigated the AZ910.5Y alloy and reported that dissolution of the β -Mg₁₇Al₁₂ phase after solid solution treatment reduced galvanic corrosion susceptibility and enhanced corrosion resistance [15]. Nevertheless, while solid solution treatment eliminates certain casting defects, it inadequately regulates the distribution of high-melting-point second phases. Overall performance enhancement is still limited.

Extrusion processing effectively refines Mg alloy grains. It fragments coarse second phase particles into smaller reinforcing phases, and simultaneously weakens the basal texture [16]. Wang et al. observed that during the extrusion of as-cast Mg1Al8Y (wt.%) alloy, thermostable Al₂Y particles further fragmented into finer, homogeneously distributed precipitates, resulting in a moderate enhancement of mechanical properties [17]. Similarly, Hu et al. demonstrated that in the hot-extruded Mg4Al2Sn0.4Mn1.0Y (wt.%) alloy, massive Al₂Y particles promoted complete DRX via PSN, concurrently weakening the basal texture and improving both mechanical properties and corrosion resistance [18]. However, the second phase fragmentations in extruded matrices tend to agglomerate. The strengthening efficacy in this path is decreased [19].

Based on the respective advantages and limitations of solution treatment and extrusion, the adoption of a “solution treatment followed by extrusion” integrated process is considered a promising Mg alloy engineering route. Solution treatment enables the dissolution of many segregated alloying elements from the as-cast alloy into the matrix through high-temperature diffusion, thereby eliminating composition-enriched zones at grain boundaries. Subsequent extrusion can achieve grain refinement, optimized distribution of second phases, and texture control on this more homogeneous matrix. However, current understanding of the evolution of microstructure and properties in solution-treated Mg-Zn-Al-Mn-Y alloys subjected to extrusion deformation remains insufficient. Crucially, systematic knowledge is lacking at Y content’s influences on the effectiveness of this integrated.

In this work, Mg_{2.0}Zn_{2.0}Al_{0.3}Mn_{*x*}Y ($0 \leq x \leq 5.0$ wt.%) alloys were designed based on non-rare-earth MgZnAlMn systems, considering Y’s solid solubility and functional mechanisms in Mg alloys. This alloy series underwent solid solution treatment com-

bined with backward extrusion processing. The effects of Y content on microstructural evolution and properties of the extruded alloys were comprehensively investigated with a systematic elucidation of the DRX mechanisms. This work is hoped to establish a concrete foundation for microstructural regulation and performance enhancement in high-performance rare-earth Mg alloys.

2. Materials and methods

High-purity raw materials were used: pure Mg (≥ 99.99 wt.%), Zn (≥ 99.99 wt.%), Al (≥ 99.9 wt.%), and the Mg-15Mn/Mg-30Y master alloys (wt.%). Mg_{2.0}Zn_{2.0}Al_{0.3}Mn_{*x*}Y alloys ($x = 0, 1.0, 2.0, 3.0, 4.0, 5.0$ wt.%) were melted in a medium-frequency induction furnace (ZGJL0.01-40-4). After holding at 800 ± 5 °C for 5 min, the melt was cooled to 760 ± 5 °C and cast into preheated steel molds (200 ± 5 °C). The as-cast ingots were machined into cylindrical specimens ($\phi 50$ mm $\times$ 35 mm), solution-treated at 520 °C for 10 h, and hot-extruded at 360 °C with a ram speed of 5 mm s^{−1} (extrusion ratio 7.7:1). Water quenching was applied on the extruded bars, followed by stress-relief annealing at 200 °C for 4 h. The extruded alloys were designated as the EX-series: E0, E1.0, E2.0, E3.0, E4.0, and E5.0, based on their Y content.

Samples were cut parallel to the extrusion direction (ED) to observe the alloy microstructure. The metallographic structure was examined using a ZEISS Axio Vert. A1 optical microscope (OM), and the grain size and distribution were quantified via the *ProImaging software*. Tensile fractures were observed and analyzed using a JSM-5610LV scanning electron microscope (SEM) equipped with an electron backscatter diffraction (EBSD) detector. EBSD samples were electrolytically polished using an electrolyte consisting of anhydrous ethanol and phosphoric acid (volume ratio 3:5). EBSD scanning was conducted after 2–5 min of polishing at 20 kV and 0.1–0.3 A at a scanning step size of 0.5 μ m. Data were processed and analyzed using *AztecCrystal software*. The alloy microstructure was examined using a JEM-2100 transmission electron microscope (TEM) equipped with an energy dispersive spectrometer (EDS) to analyze micro-area composition and phase structure. The three-dimensional morphology of the alloy surface after corrosion was observed using a SENSOFAR S NEOX 3D confocal white light interference profilometer. Room-temperature tensile tests were conducted on an AG-1250 KN material testing machine. Tensile specimens had a gauge dimension of 15 mm $\times$ 3.1 mm $\times$ 2 mm, with a tensile rate of 1 mm min^{−1}. Three parallel specimens were prepared for each group.

An Autolab PGSTAT128 N electrochemical workstation was employed. It was equipped with a three-

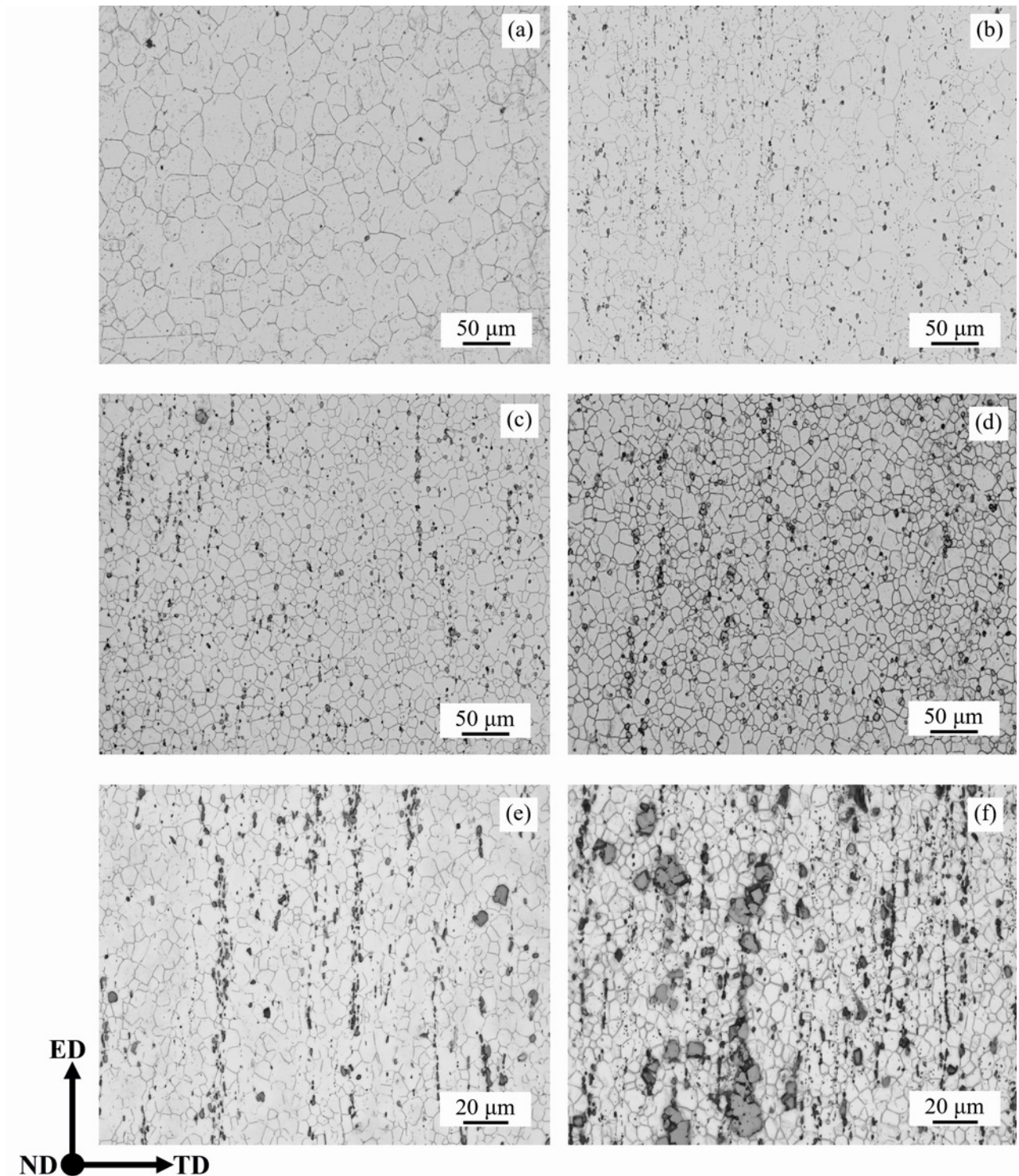


Fig. 1. Optical microstructure images of as-extruded $\text{Mg}_{2.0}\text{Zn}_{2.0}\text{Al}_{0.3}\text{Mn}_x\text{Y}$ alloys along the ED (To clearly display the microstructure, Figs. 1e and 1f were taken at a higher magnification ($500\times$)): (a) E0, (b) E1.0, (c) E2.0, (d) E3.0, (e) E4.0, (f) E5.0.

-electrode system and operated in a 3.5% NaCl solution at 25 °C. The reference electrode was a saturated calomel electrode (SCE), the counter electrode a high-density graphite sheet, and the working electrode a

cylindrical sample with a working area of 1 cm^2 . Polarization curves were scanned from -0.25 to $+0.4\text{ V}$ relative to the open circuit potential (OCP) at a scan rate of 1 mV s^{-1} .

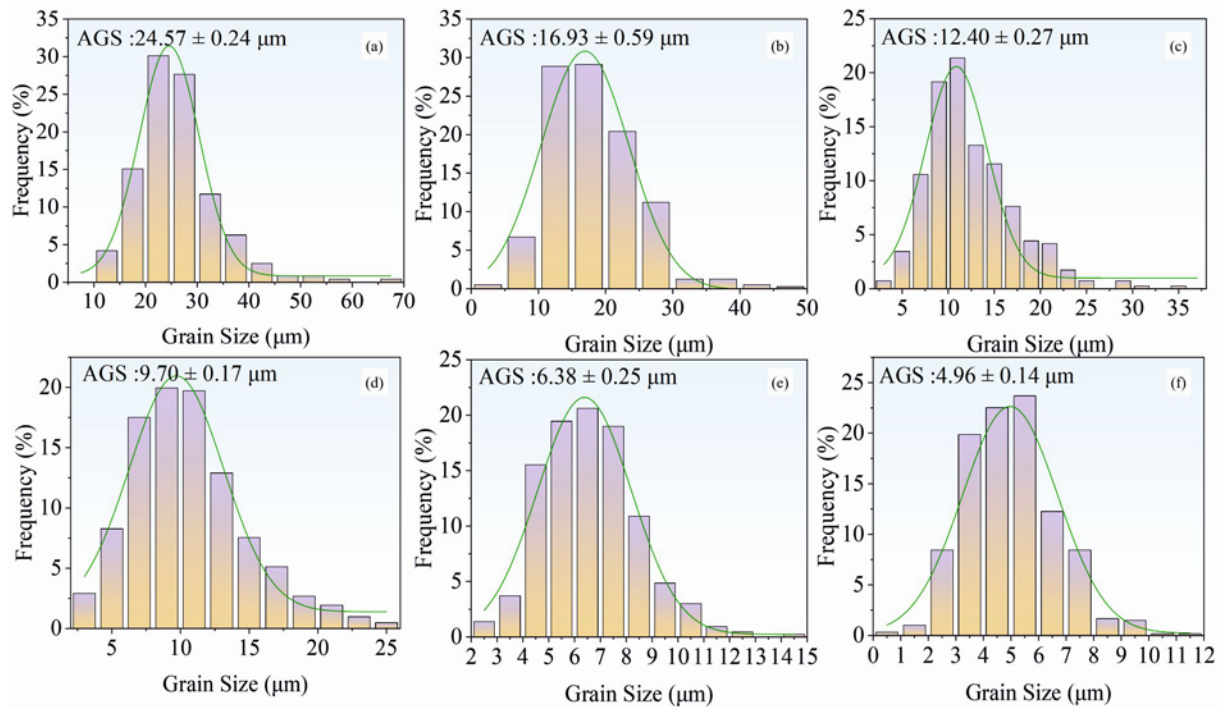


Fig. 2. Grain size frequency distribution map of as-extruded $\text{Mg}_{2.0}\text{Zn}_{2.0}\text{Al}_{0.3}\text{Mn}_x\text{Y}$ alloys: (a) E0, (b) E1.0, (c) E2.0, (d) E3.0, (e) E4.0, and (f) E5.0.

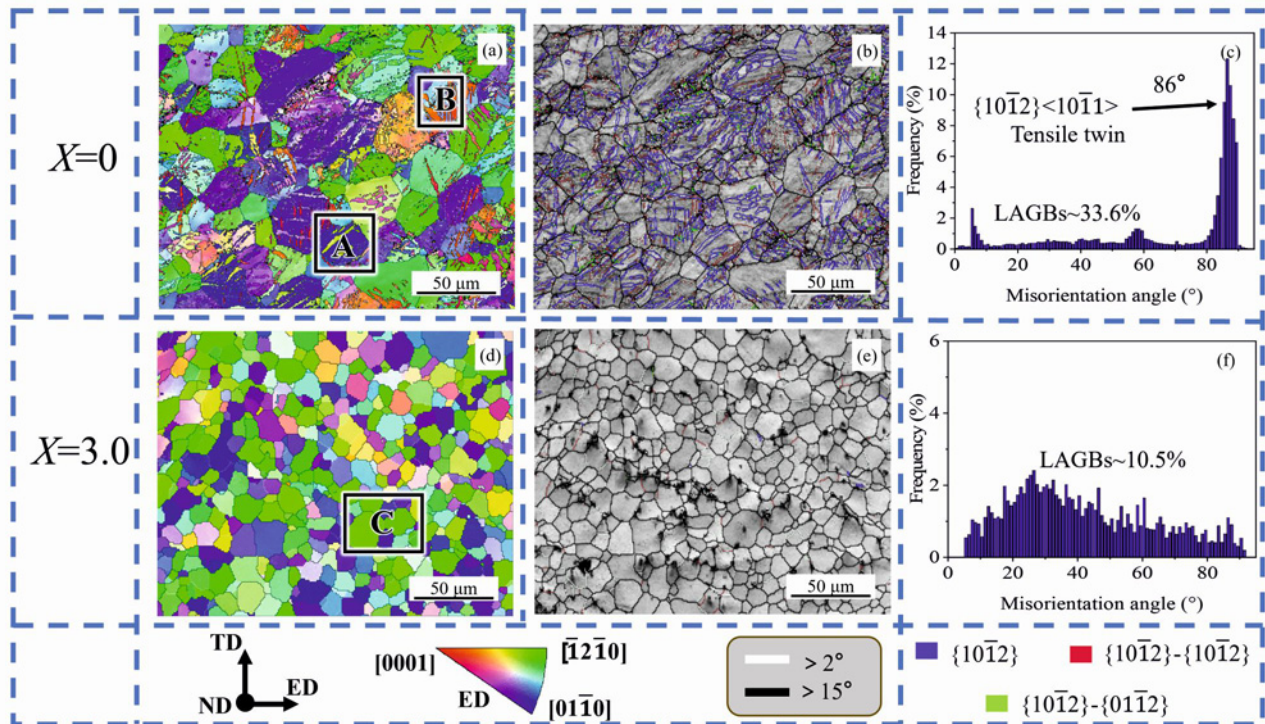


Fig. 3. EBSD analysis of E0 and E3.0 alloys: (a) inverse pole figure (IPF) map of E0, (b) grain boundary map of E0 (in these maps, white lines denote LAGBs with orientation deviations of 2° – 15°), black lines denote HAGBs (with orientation deviations $>15^\circ$), blue lines denote $\{10\bar{1}2\}$ tensile twin boundaries, red lines denote $\{10\bar{1}2\}$ – $\{10\bar{1}2\}$ secondary tensile twin boundaries, and green lines denote $\{10\bar{1}2\}$ – $\{01\bar{1}2\}$ secondary tensile twin boundaries), (c) misorientation angle distribution of E0, (d) IPF map of E3.0, (e) grain boundary map of E3.0, and (f) misorientation angle distribution of E3.0. and angle distributions map of the E3.0 alloy, respectively.

Table 1. Element contents (at.%) of the second phase marked in Fig. 6

Position	Mg	Zn	Al	Mn	Y
A	5.81	0.57	62.41	0.90	30.31
B	7.18	8.67	65.78	0.08	18.29
C	9.04	0.53	58.65	0.25	31.53
D	3.53	9.12	65.12	0.31	21.92

Weight loss corrosion tests and hydrogen evolution corrosion tests were conducted in a 3.5% NaCl solution at 25 °C for 72 h. The solution was totally replaced every 8 h. The ratio of sample surface area (cm²) to solution volume (mL) was maintained at approximately 1:30. A funneltitration tube system was used for the hydrogen evolution tests to collect gas and record the hydrogen evolution volume. The calculation methods for the weight loss corrosion rate (CR_w in mm y⁻¹) and hydrogen evolution corrosion rate (CR_H in mm y⁻¹) are provided in Eqs (1) and (2) [20, 21]:

$$CR_w = 87.6\Delta W/(\rho At), \quad (1)$$

$$CR_H = 95.36\Delta V/(\rho At), \quad (2)$$

where ΔW is the weight loss (mg) of the sample after corrosion, ρ is sample density (g cm⁻³), A is the surface area (cm²) of the sample exposed to the solution, t is immersion corrosion time (h), and ΔV is the volume of collected hydrogen (mL). The average value was calculated for the three sample groups.

3. Results

3.1. Microstructures

Figures 1 and 2 present the optical microstructures and corresponding grain size distributions of the as-extruded Mg-2.0Zn-2.0Al-0.3Mn- x Y alloys along the ED, respectively. The average grain size exhibits a progressive reduction with increasing Y content. The measured average grain sizes of the E0, E1.0, E2.0, E3.0, E4.0, and E5.0 alloys are $24.57 \pm 0.24 \mu\text{m}$, $16.93 \pm 0.59 \mu\text{m}$, $12.40 \pm 0.27 \mu\text{m}$, $9.70 \pm 0.17 \mu\text{m}$, $6.38 \pm 0.25 \mu\text{m}$, and $4.96 \pm 0.14 \mu\text{m}$, respectively. In the Y-free alloy (E0), alloying elements are predominantly dissolved within the matrix as observed in Fig. 1a. As the Y content increases from 1.0 to 3.0 wt.% (Figs. 1b–d), black second phase particles gradually emerge and distribute discretely along the ED, with both their volume fraction and particle size increasing monotonically. From Figs. 1e,f, at 4.0–5.0 wt.% Y, the quantity and size of black second phase particles increase significantly, with these particles continuing to be distributed along the ED.

Figure 3 presents the inverse pole figure (IPF) maps, grain boundary character distribution maps, and misorientation angle distribution maps of the E0 and E3.0 alloys. As shown in Figs. 3a,b, numerous low-angle grain boundaries (LAGBs), $\{10\bar{1}2\}$ tensile twins, $\{10\bar{1}2\}$ – $\{10\bar{1}2\}$ secondary tensile twins, and $\{10\bar{1}2\}$ – $\{01\bar{1}2\}$ secondary tensile twin boundaries are formed in the alloy without Y element after hot extrusion. Secondary tensile twinning typically develops on the basis of $\{10\bar{1}2\}$ tensile twinning, and its formation is related to local stress accommodation and strain incompatibility. Additionally, Fig. 3c shows that the intensity of the misorientation angle distribution is significantly higher at approximately 86° than in other angular regions, which corresponds well with the abundant $\{10\bar{1}2\}$ tensile twins present in the alloy. From Figs. 3d,e, when the Y content is 3.0 wt.%, the number of tensile twins decreases significantly. This indicates that Y addition effectively suppresses twinning, transforming the deformation mechanism from a twindominated plastic deformation mode to a slipdominated one. Figure 3f shows that the misorientation angle distribution is primarily concentrated within the 20°–40° range, with the proportion of LAGBs in the alloy reduced to approximately 10.5 %. This indicates that the grains of the alloy under this composition are predominantly high-angle grain boundaries (HAGBs).

Figure 4 shows the Kernel Average Misorientation (KAM) distribution maps and statistical histograms of the E0 and E3.0 alloys. Generally, the KAM value enables quantitatively characterize the degree of local lattice distortion, thereby evaluating the magnitude of stored deformation energy and the density of geometrically necessary dislocations (GNDs) [22]. From Figs. 4a,b, the average KAM value of the E0 alloy is approximately 1.1°, with high-density dislocation regions concentrated at grain boundaries and twin boundaries. Figures 4c,d illustrate that the average KAM value of the E3.0 alloy is approximately 0.38°, representing a significant decrease in dislocation density relative to the E0 alloy. This indicates that Y addition enhances dislocation annihilation during DRX, substantially reducing the stored deformation energy of the alloy, which is consistent with the decline in LAGBs in Fig. 3f.

Figure 5 presents the (0001) pole figures (PF) and corresponding antipole figures of the E0 and E3.0 al-

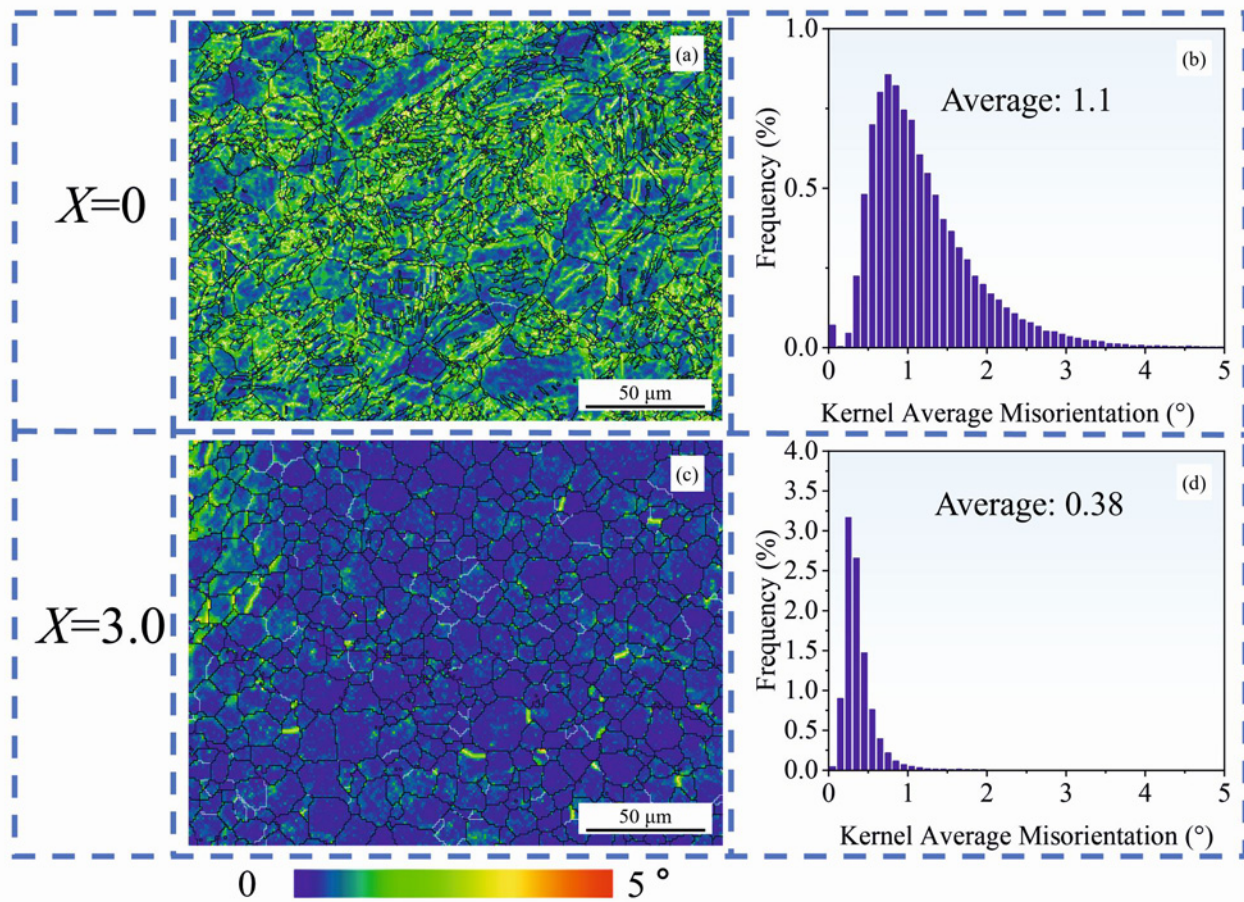


Fig. 4. KAM maps and distributions for E0 and E3.0 alloys: (a) KAM map of E0 alloy, (b) average KAM value distribution of E0 alloy, (c) KAM map of E3.0 alloy, and (d) average KAM value distribution of E3.0 alloy.

loys. As shown in the figure, grains in both alloys exhibit a certain preferred orientation, characteristic of a typical (0001) basal texture. As illustrated in Figs. 5a,b, the maximum basal texture intensity of the E0 alloy is approximately 7.88 mrd, with primary textures consisting of $\langle 01\bar{1}0 \rangle // \text{ED}$ and $\langle \bar{1}2\bar{1}0 \rangle // \text{ED}$ fibers. The maximum basal texture intensity of the E3.0 alloy is approximately 12.59 mrd. This indicates that the addition of Y significantly increases the basal texture strength of the alloy, even though $\langle 01\bar{1}0 \rangle // \text{ED}$ and $\langle \bar{1}2\bar{1}0 \rangle // \text{ED}$ still dominate the texture orientation. The increase in basal texture intensity is primarily associated with the reduction in twin density, which promotes the preferential rotation of grains toward the $\langle 0001 \rangle // \text{ND}$ direction during DRX.

Typical TEM images are depicted in Fig. 6 for the E3.0 and E5.0 alloys with the selected area electron diffraction (SAED) patterns. Element composition data obtained from corresponding EDS analysis of particles A, B, C, and D are summarized in Table 1. As observed in Figs. 6a,b, nanosized second phase particles are distributed both at grain boundaries and

within grain interiors. The SAED pattern of particle A shown in Fig. 6b indicates that these second phase particles possess a face-centered cubic structure. The interplanar spacings of 0.2775, 0.1801, and 0.1329 nm in the SAED patterns correspond to the standard interplanar spacings of 0.2779, 0.1803, and 0.1329 nm for the (202), (133), and (135) crystal planes of the Al_2Y phase. Furthermore, Table 1 shows that the particles are mainly composed of Al and Y, with an Al:Y atomic ratio near 2:1, which verifies that the precipitates are Al_2Y [23]. From Fig. 6c, particle B is located at the grain boundary and exhibits a fragmented morphology. This indicates that these second phase particles result from the fragmentation of larger undissolved second phase particles during the extrusion process. SAED pattern indexing of particle B identifies a bodycentered orthorhombic structure. The measured interplanar spacings of 0.2023, 0.2656, and 0.1613 nm in the SAED patterns are consistent with the standard values of 0.2039, 0.2644, and 0.1639 nm for the (211), (042), and (233) crystal planes of the Al_{11}Y_3 phase, respectively. Likewise, Table 1 shows that the particle is principally composed of Al, Zn, and Y, with

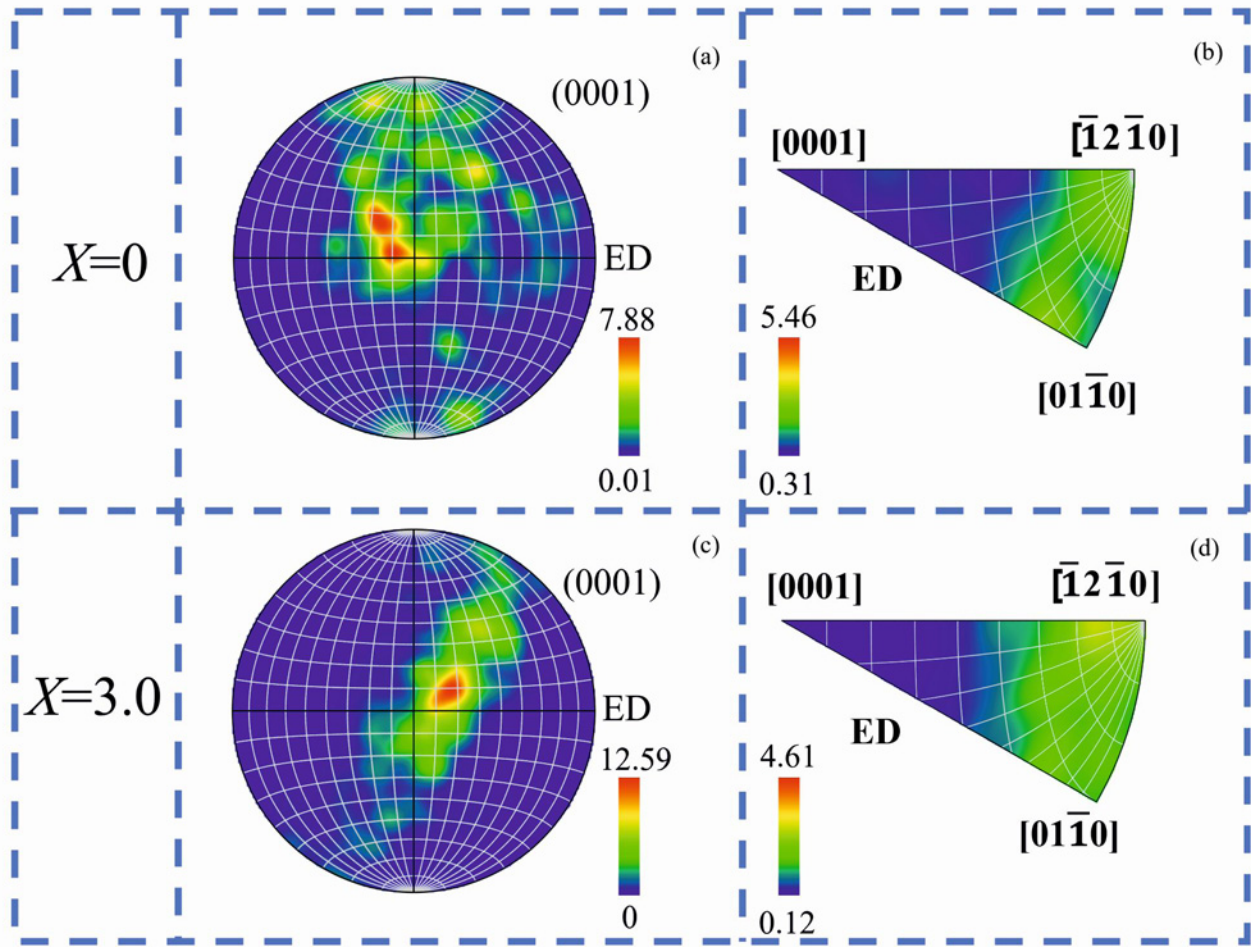


Fig. 5. (0001) pole figures and inverse pole figures for E0 and E3.0 alloys: (a) pole figure of E0, (b) inverse pole figure of E0, (c) pole figure of E3.0, and (d) inverse pole figure of E3.0.

Table 2. Mechanical properties of as-extruded Mg_{2.0}Zn_{2.0}Al_{0.3}Mn_xY alloys

Materials	E0	E1.0	E2.0	E3.0	E4.0	E5.0
UTS (MPa)	239.6 ± 2.3	256.6 ± 2.3	273.6 ± 2.4	289.4 ± 2.8	307.7 ± 2.8	316.65 ± 1.5
YS (MPa)	153.3 ± 2.5	169.61 ± 2.2	178.7 ± 2.9	214.1 ± 2.1	283.1 ± 2.3	267.95 ± 1.7
EL (%)	18.1 ± 0.2	19.6 ± 0.3	21.8 ± 0.4	25.8 ± 0.5	19.2 ± 0.4	16.4 ± 0.3

the ratio of total Al and Zn atoms to Y atoms close to 11:3, confirming that the particle is (Al, Zn)₁₁Y₃ [24].

The second phase particles with an average size of approximately 1 μm can be observed to hinder dislocations in Fig. 6d. Figure 6e shows that coarse second phase particles are more concentrated. SAED pattern calibration of particle C in Fig. 6e, combined with EDS analysis (Table 1), it is confirmed that this particle is Al₂Y, identical to particle A presented in Fig. 6b. These particles are likely larger Al₂Y particles in the alloy prior to extrusion, and get fragmented during extrusion and remain in the matrix as agglomerates. From Fig. 6f, particle D with a length of 5 μm is observed similar to the fractured particle B. SAED cal-

ibration of particle D, combined with EDS analysis (Table 1), verifies that this particle, similar to particle B illustrated in Fig. 6c, is identified as the (Al, Zn)₁₁Y₃.

3.2. Mechanical properties

Figure 7 and Table 2 present the engineering stress-strain curves of as-extruded Mg-2.0Zn-2.0Al-0.3Mn-*x*Y alloys at room temperature, along with their corresponding ultimate tensile strength (UTS), yield strength (YS), and elongation (EL) values. As shown, UTS and YS of the alloys exhibit a gradual increase with increasing Y content, while EL first increases and subsequently decreases. At a Y content

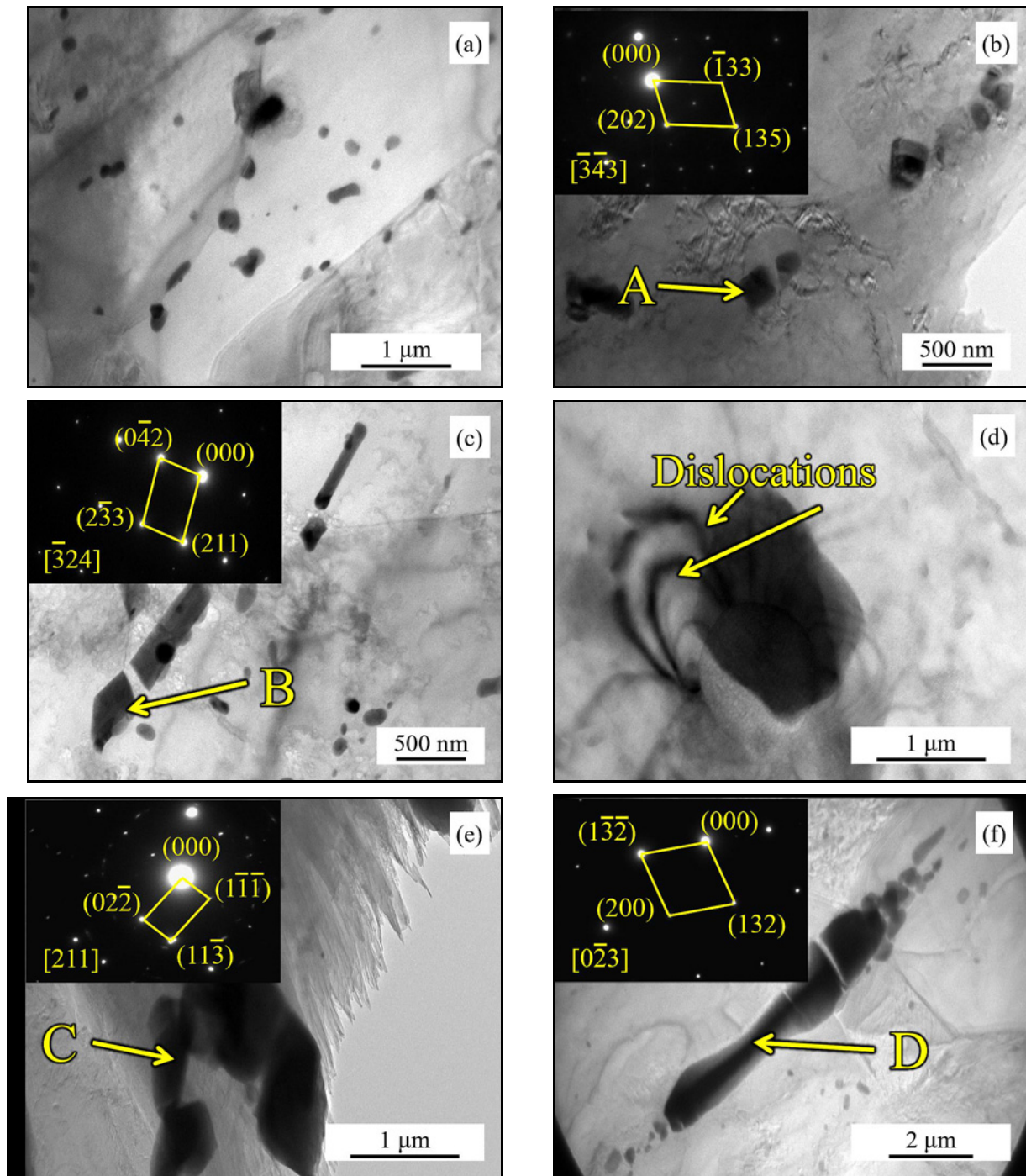


Fig. 6. TEM bright-field images and SAED patterns of as-extruded E3.0 and E5.0 alloys: (a)–(c) E3.0 alloy showing nanoscale second phase particles (Al_2Y and $(\text{Al,Zn})_{11}\text{Y}_3$) and (d)–(f) E5.0 alloy showing coarse second phase particles.

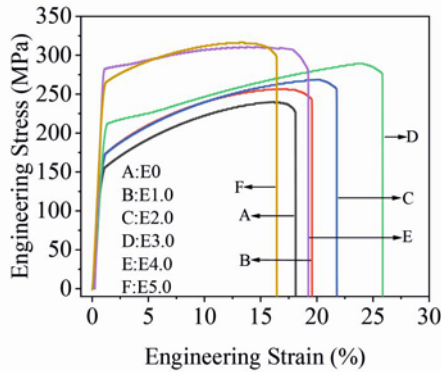
of 3.0 wt.%, the alloy achieves an excellent balance between strength and plasticity, with UTS, YS, and EL reaching 289.4 ± 2.8 MPa, 214.1 ± 2.1 MPa, and 25.8 ± 0.5 %, respectively.

Figure 8 illustrates the tensile fracture morphologies of as-extruded $\text{Mg}_{2.0}\text{Zn}_{2.0}\text{Al}_{0.3}\text{Mn}_x\text{Y}$ alloys. The fracture surfaces are characterized by cleavage planes and dimples with varying densities and dimensions, exhibiting a hybrid fracture mode of cleavage and

ductile dimple fracture. At 0 wt.% Y, the tensile fracture contains large cleavage planes, few dimples, and small microcracks. At 1.0–3.0 wt.% Y, with increasing Y content, the density and size of the dimples increase progressively, while the dimensions and abundance of cleavage planes diminish accordingly. At 3.0 wt.% Y, the tensile fracture contains many dimples with significant size, indicating good EL of the alloy, in line with observation from Fig. 7. At 4.0–5.0 wt.% Y, the

Table 3. Tafel fitting results of as-extruded Mg_{2.0}Zn_{2.0}Al_{0.3}Mn_xY alloys

Alloys	E_{corr} (V)	I_{corr} ($\mu\text{A cm}^{-2}$)	P_i (mm y^{-1})
E0	-1.465 ± 0.011	9.380 ± 0.013	0.214 ± 0.005
E1.0	-1.401 ± 0.012	4.202 ± 0.012	0.096 ± 0.004
E2.0	-1.392 ± 0.011	3.852 ± 0.011	0.088 ± 0.003
E3.0	-1.313 ± 0.013	2.449 ± 0.015	0.055 ± 0.005
E4.0	-1.437 ± 0.012	5.732 ± 0.012	0.131 ± 0.005
E5.0	-1.452 ± 0.015	8.722 ± 0.014	0.199 ± 0.004

Fig. 7. Engineering stress-strain curves of as-extruded Mg_{2.0}Zn_{2.0}Al_{0.3}Mn_xY alloys at room temperature.

number of coarse second phase particles on the fracture surface increases significantly, and the number and size of dimples undergo a gradual reduction; Furthermore, microvoids are observed to nucleate and grow around these coarse second phase particles. The formation of microvoids are primarily associated with stress concentration induced fracture of brittle second phase particles and interfacial debonding caused by deformation incompatibility between the matrix and second phase particles [25].

Figure 9 presents the polarization curves of as-extruded Mg_{2.0}Zn_{2.0}Al_{0.3}Mn_xY alloys following 1 h of immersion in a 3.5% NaCl solution. The corrosion potential (E_{corr}), corrosion current density (I_{corr}), and corrosion rate (P_i) were determined via Tafel extrapolation based on Faraday's law and are summarized in Table 3. The relationship between P_i and I_{corr} is expressed as $P_i = 22.85 I_{\text{corr}}$ [26, 27]. As shown in the figure, with increasing Y content, E_{corr} exhibits an initial positive shift followed by a negative shift, while I_{corr} and P_i first decrease and subsequently increase. The 3.0 wt.% alloy has a relatively positive E_{corr} (-1.313 ± 0.013 V), a low I_{corr} ($2.449 \pm 0.015 \mu\text{A cm}^{-2}$), and a low P_i ($0.055 \pm 0.05 \text{ mm y}^{-1}$). These values confirm the excellent corrosion resistance of the E3.0 alloy.

Figure 10 presents the CR_w and CR_H of as-extruded Mg_{2.0}Zn_{2.0}Al_{0.3}Mn_xY alloys after 72 h of immersion in a 3.5% NaCl solution. As observed,

with increasing Y content, both CR_w and CR_H first decrease and then increase. At 3.0 wt.% Y, the alloy exhibits a low corrosion rate, with $CR_w = 3.31 \pm 0.16 \text{ mm y}^{-1}$ and $CR_H = 3.02 \pm 0.13 \text{ mm y}^{-1}$.

Figure 11 presents the three-dimensional corrosion morphologies of as-extruded Mg_{2.0}Zn_{2.0}Al_{0.3}Mn_xY alloys after the removal of corrosion products. Surface roughness (Sa) and line roughness (Ra) across localised corrosion regions can partially reflect the corrosion resistance of the alloys. Lower Sa and Ra values correspond to superior corrosion resistance [28]. Without Y, Sa is determined to be $7.95 \mu\text{m}$ and Ra to be $10.11 \mu\text{m}$, as shown in Fig. 11a. Large corrosion pits appear on the surface indicating severe localised corrosion. At 1.0–3.0 wt.% Y (Figs. 11b–d), increasing Y content leads to a gradual decrease in Sa and Ra , accompanied by a transition from localised to uniform corrosion and a reduction in corrosion pit dimensions. At 4.0–5.0 wt.% Y (Figs. 11e,f), Sa and Ra increase progressively with increasing Y content. Corrosion pits increase in depth, quantity, and diameter, displaying a continuous streamline distribution along the ED.

4. Discussions

4.1. Microstructure evolution

As evidenced in Figs. 1–3, the addition of Y to the Mg_{2.0}Zn_{2.0}Al_{0.3}Mn_xY alloys significantly refines the grain structure and suppresses twinning activity. The E0 alloy exhibits the largest average grain size, with a high density of tensile twins distributed in the matrix. To analyze the twinning behavior of the E0 alloy during the extrusion process, two representative regions A and B were selected from Fig. 3a for analysis.

Figure 12 presents the EBSD results and analysis of region A in Fig. 3a. To facilitate the investigation of intragranular twinning behavior and grain orientation evolution, a black arrow (A to B) is marked within the grains in Fig. 12a. As shown in Figs. 12a,d, and e, the parent grain M exhibits distinct basal texture characteristics, with the orientations of twins T1 and T2 differing significantly from that of M. Thus, twin formation effectively reduces basal texture inten-

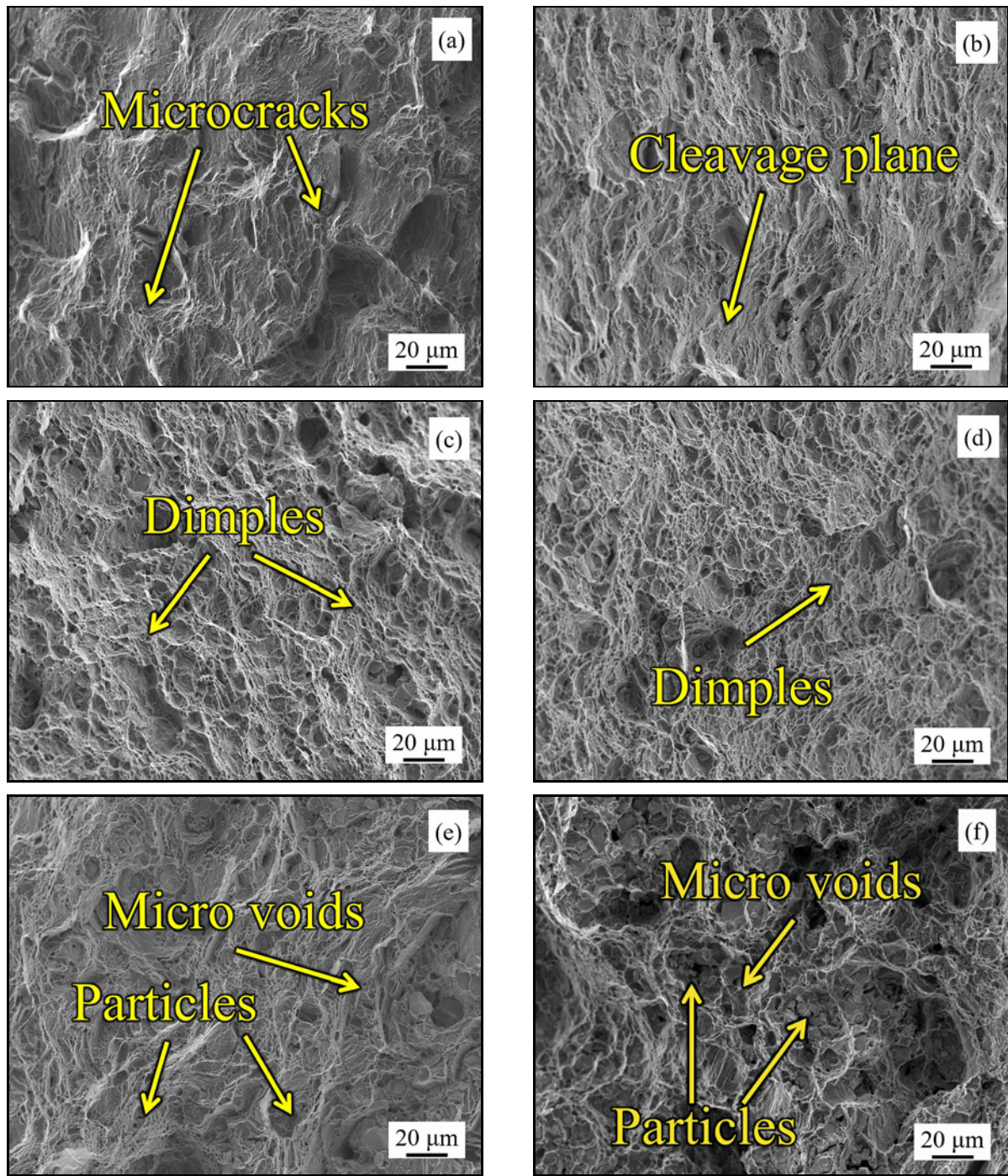


Fig. 8. SEM images of tensile fracture surfaces of as-extruded $\text{Mg}_{2.0}\text{Zn}_{2.0}\text{Al}_{0.3}\text{Mn}_x\text{Y}$ alloys: (a) E0, (b) E1.0, (c) E2.0, (d) E3.0, (e) E4.0, and (f) E5.0.

sity and orientation concentration. Notably, KAM values at grain boundaries and within twins are considerably higher than those within the parent grain M. This may be attributed to substantial plastic deformation during twin formation, as well as high dislocation density and stress concentration (Fig. 12c). Additionally, the misorientation distribution map along the AB direction (Fig. 12f) reveals four misorientation peaks at approximately 86° , corresponding to $\{10\bar{1}2\}$

tensile twin boundaries – a typical feature of such boundaries. In Mg alloys, $\{10\bar{1}2\}$ tensile twins exhibit low shear strain and low critical resolved shear stress (CRSS). Consequently, $\{10\bar{1}2\}$ tensile twins are preferentially formed during extrusion deformation. Additionally, stress concentration regions formed by dislocation pileup or interface blocking at $\{10\bar{1}2\}$ twin boundaries provide the driving force for secondary twins. From Fig. 12b, the formation of $\{10\bar{1}2\}$ – $\{10\bar{1}2\}$

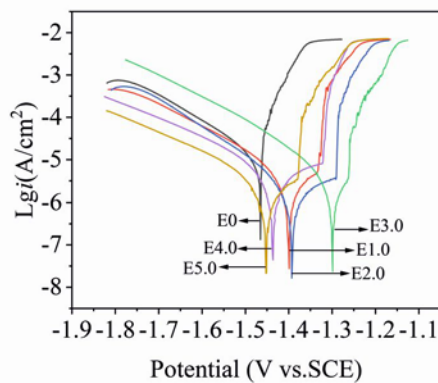


Fig. 9. Polarization curves of as-extruded Mg_{2.0}Zn_{2.0}Al_{0.3}-Mn_xY alloys after 1 h immersion in 3.5% NaCl solution.

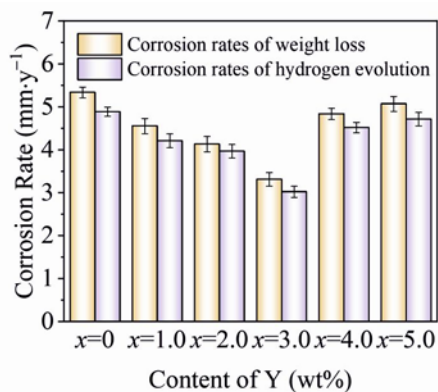


Fig. 10. Average value CR_w and CR_H of as-extruded Mg_{2.0}Zn_{2.0}Al_{0.3}Mn_xY alloys after 72 h immersion in 3.5% NaCl solution.

and $\{10\bar{1}2\}$ – $\{01\bar{1}2\}$ secondary twins can be observed. Moreover, distortion induced by dislocation pileup at twin boundaries supplies substantial energy for the occurrence of DRX [29]. The primary role of these tensile twins is to work with dislocation slip to accommodate local strain and alleviate stress concentration, thereby enhancing the alloy's plastic deformability [30].

Figure 13 presents the EBSD analysis results of region B in Fig. 3a and corresponding typical TEM images of the E0 alloy. To further analyze the DRX behavior of the E0 alloy during extrusion process, the parent grain M in this region was marked with lines AB and CD. Furthermore, grains (G1, G2) and a twin (T1) within the parent grain M were selected, with the corresponding three-dimensional orientation maps. From Figs. 13a–c, grain G1 is located within the parent grain M, surrounded by a high density of LAGBs. It is a characteristic of the CDRX mechanism [31]. Moreover, the AB orientation difference distribution map (Fig. 13d) shows that grain G1 has formed

HAGBs with an orientation difference exceeding than 15° . In general, at high temperatures and high strain rates, dislocations can form subgrain boundaries via dislocation climb and crossslip. Subsequently, these subgrain boundaries gradually evolve into HAGBs via dislocation climb. Which reduces the system's free energy and facilitates the formation of new DRXed grains [32]. The newly formed equiaxed recrystallized grain G1 exhibits low dislocation density, indicating that dislocations have been gradually absorbed via grain boundary migration. Finally, the CDRX occurs via structural reconstruction through subgrain evolution.

Twin formation exerts a dual benefits effect on microstructure refinement. First, it provides new paths for dislocation motion, thereby significantly enhancing the alloy's plastic deformability. Second, the formation effectively lowers the energy barrier for DRX nucleation and grain growth. Thus, TDRX is another predominant recrystallisation mechanism in Mg alloys, whose core processes involve twin formation, the transition of twin boundaries to random boundaries, and the growth of new grains. As shown in Figs. 13c,g,h, high-density dislocations are typically present within and adjacent to twins. These dislocation accumulations supply sufficient energy to induce DRX nucleation. Combined with Figs. 13a,e, HAGBs and twin boundaries coexist around grain G2, indicating that G2 nucleation is dependent on the TDRX mechanism. As high-distortion-energy interfaces, twin boundaries can either serve directly as recrystallisation nucleation sites or, through interactions between dislocation cells and twin boundaries, promote subgrain rotation and merging to form new HAGBs, which eventually grow into equiaxed DRX grains. Thus, at 0 wt.% Y, the synergistic effect of CDRX and TDRX governs the alloy's postextrusion recrystallized structure. CDRX dominates continuous refinement in steady-state deformation zones, while TDRX triggers DRX in twin-concentrated regions. These mechanisms interact synergistically to shape the alloy's complex recrystallized texture and the morphology of dynamic recrystallized grains.

At 1.0–5.0 wt.% Y, the addition of Y significantly modifies the microstructure evolution mechanism during alloy extrusion, leading to a progressive reduction in the alloy's average grain size. The size and quantity of second phase particles increase with increasing Y content, while twin formation is effectively suppressed. To analyze the DRX mechanism of Y-containing alloys during hot extrusion, region C was selected from Fig. 3d. The TEM image of the E3.0 alloy is provided in Fig. 14. From Fig. 14a, the boundaries of grains G1 and G2 exhibit a bulging morphology, leading to the formation of grain G3, which is one of the typical characteristics of DDRX. This phenomenon is primarily associated with the deformation energy storage gradi-

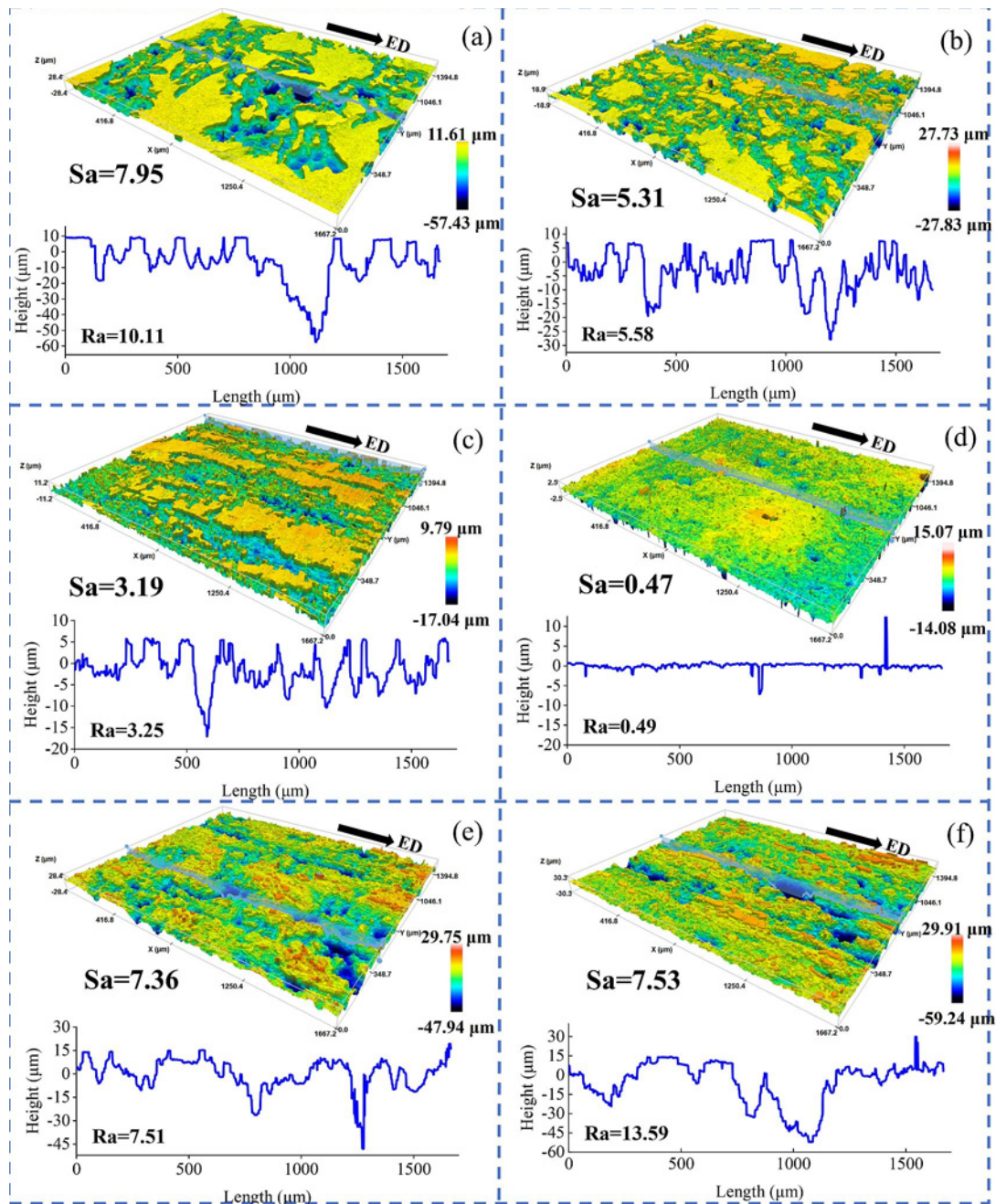


Fig. 11. Three-dimensional corrosion morphology of as-extruded Mg_{2.0}Zn_{2.0}Al_{0.3}Mn_xY alloys after 72 h of immersion in 3.5% NaCl solution to remove the corrosion products: (a) E0, (b) E1.0, (c) E2.0, (d) E3.0, (e) E4.0, and (f) E5.0.

ent (i.e., dislocation density difference) between adjacent grains. The high-energy-storage grain boundaries release energy via boundary bulging, accelerate interface migration, and drive DDRX nucleation. Additionally, from Figs. 14b,c, the orientation of grain G3 differs significantly from those of grains G1 and G2, indicating that the orientation of grains formed via DDRX nucleation is independent of the original grains. This is conducive for weakening the alloy's basal texture and reducing its anisotropy. Furthermore, the dislocation cells in Figs. 14d,e are representative feature

from the CDRX mechanism. During extrusion, accumulated shear strain induces local residual dislocations, which entangle, proliferate, and reorganize to form dislocation cells. These cells first undergo dynamic recovery, transforming into subgrains bounded by LAGBs. Subgrains with LAGBs then evolve into DRX grains with HAGBs through continuous dislocation absorption.

During hot extrusion, the size and quantity of second phase particles regulate DRX and grain refinement via PSN and the Smith-Zener pinning effect.

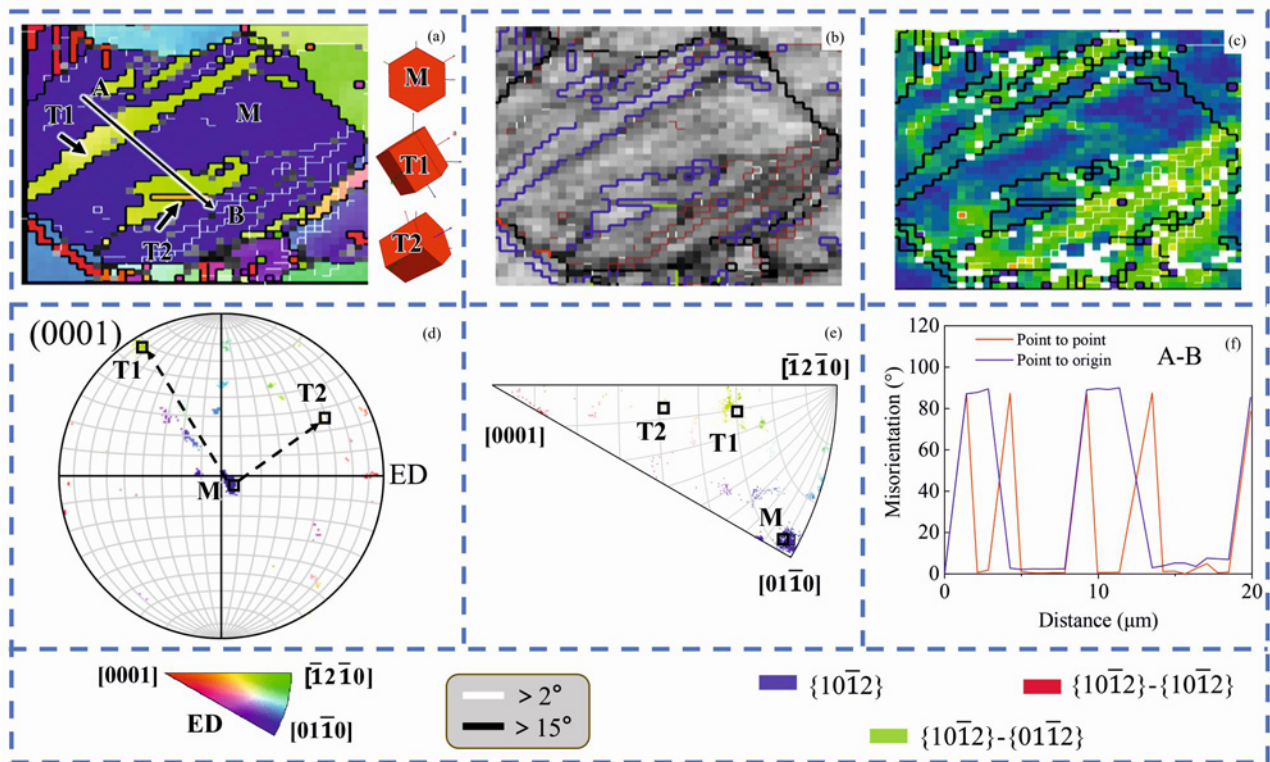


Fig. 12. EBSD analysis of region A in E0 alloy (from Fig. 3a): (a) IPF map and 3D orientation diagram, (b) band contrast map, (c) KAM map, (d) (0001) pole figure, (e) inverse pole figure, and (f) misorientation profile along line AB.

Dislocations and shear bands tend to accumulate at the vicinity of second phase particles larger than 1 μm , inducing lattice distortion and orientation gradients. This provides additional nucleation sites for DRX and benefits the PSN mechanism. Nanoscale second phase particles smaller than 1 μm are distributed at grain boundaries and within grains interiors. They can pin grain boundaries, hinder grain boundary sliding, inhibit grain growth, and delay the DRX process [33, 34]. Thus, as observed in Fig. 6, nanoscale second phase particles dispersed within grains and at grain boundaries enhance the Smith-Zener pinning effect, while larger micron-scale second phase particles facilitate the PSN mechanism. With increasing Y content, the increasing size and quantity of second phase particles further enhance DRX nucleation. Therefore, the addition of rare earth Y significantly promotes DRX and reduces the stored deformation energy of the alloy. Under the combined action of DDRX, CDRX, PSN, and the Smith-Zener pinning effect, the alloy's grains are continuously refined.

Introducing Y into the alloy system exerts a significant influence on twin formation. Y reduces the CRSS of non-basal slip systems, facilitating slip-dominated plastic deformation and thereby reducing the need for twin formation. Consequently, the addition of Y effectively suppresses twin formation. Notably, twins induce a shift in grain orientations toward non-basal directions. Y-induced suppression of twins diminishes

this orientation randomization, allowing a greater number of grains to retain or adopt the $\langle 0001 \rangle // \text{ND}$ orientation during deformation, which strengthens the basal texture.

Figure 15 illustrates the DRX mechanism of extruded $\text{Mg}_{2.0}\text{Zn}_{2.0}\text{Al}_{0.3}\text{Mn}_x\text{Y}$ alloys. In the absence of Y, the postextrusion microstructure contains numerous twins accompanied by extensive dislocation accumulation. Here, TDRX dominates as shown by twin boundaries which act as high-distortion-energy regions to induce dislocation entanglement and subgrain formation, thereby promoting DRX. With Y addition, the number of twins in the hotextruded microstructure is significantly reduced. DDRX, CDRX, and PSN synergistically drive DRX progression, ultimately achieving grain refinement and microstructural homogenization.

4.2. Microstructure influence on mechanical properties

As shown in Figs. 1 and 7, the addition of Y significantly refines the microstructure of as-extruded $\text{Mg}_{-2.0}\text{Zn}_{-2.0}\text{Al}_{-0.3}\text{Mn}_{-x}\text{Y}$ alloys and improves the alloy's mechanical properties to varying extents. In general, the mechanical properties of alloys are closely related to grain size, dislocation density, and second phase characteristics (type, size, and distribution).

Due to the high Hall-Petch coefficient of Mg al-

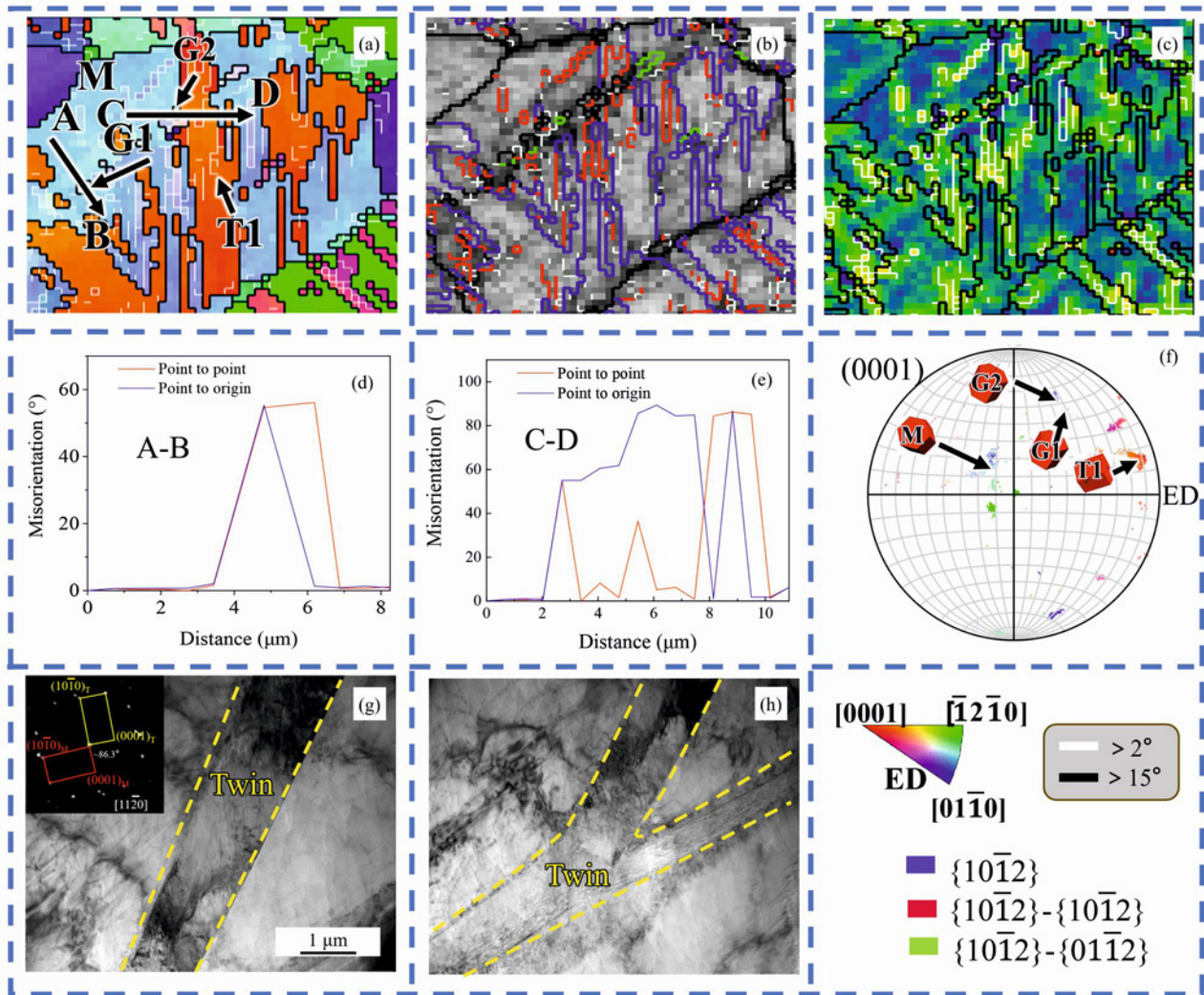


Fig. 13. EBSD analysis of region B in E0 alloy and corresponding TEM images: (a) IPF map; (b) band contrast map; (c) KAM map, (d) and (e) misorientation profiles along lines AB and CD, (f) pole figure and 3D orientation map, (g) and (h) TEM images showing dislocation structures.

loys, variations in grain size exert a pronounced influence on mechanical properties [35]. Larger grain sizes reduce the grain boundary area per unit volume, impairing the ability of grain boundaries to coordinate plastic deformation. According to the Hall-Petch relationship, the E0 alloy with a larger grain size ($24.57 \pm 0.24 \mu\text{m}$) exhibits an inferior grain boundary strengthening effect. From Figs. 3a and 4a, despite numerous $\{10\bar{1}2\}$ tensile twins contribute to partial strain accommodation, extensive dislocation pile-ups are present in twin boundary regions. During tensile deformation, stress concentration typically arises in dislocation pile-up regions, leading to the formation of microvoids or early microcracks, thereby deteriorating the plasticity of the alloy [36]. Consequently, the positive effect of twin-coordinated deformation is ultimately counteracted by interface-induced stress concentration and subsequent microdamage, resulting in

the inferior mechanical properties of the E0 alloy.

From Figs. 1b–d and Fig. 3d, at 1.0–3.0 wt.% Y, increasing Y content leads to continuous grain refinement and significant suppression of twin formation. A reduced grain size increases the grain boundary density per unit volume and the energy required for boundaries to impede dislocation motion. It also shortens the dislocation slip distance within grains during plastic deformation, thereby enhancing the capacity to accommodate plastic deformation between adjacent grains and mitigating stress concentration and microcrack formation at grain boundaries [37]. As shown in Figs. 1 and 6, a high density of nanoscale second phase particles are dispersed in the E3.0 alloy. These particles typically form a semi-coherent interface with the α -Mg matrix, exhibiting good lattice matching. Such favorable lattice matching facilitates continuous dislocation slip, thereby enhancing

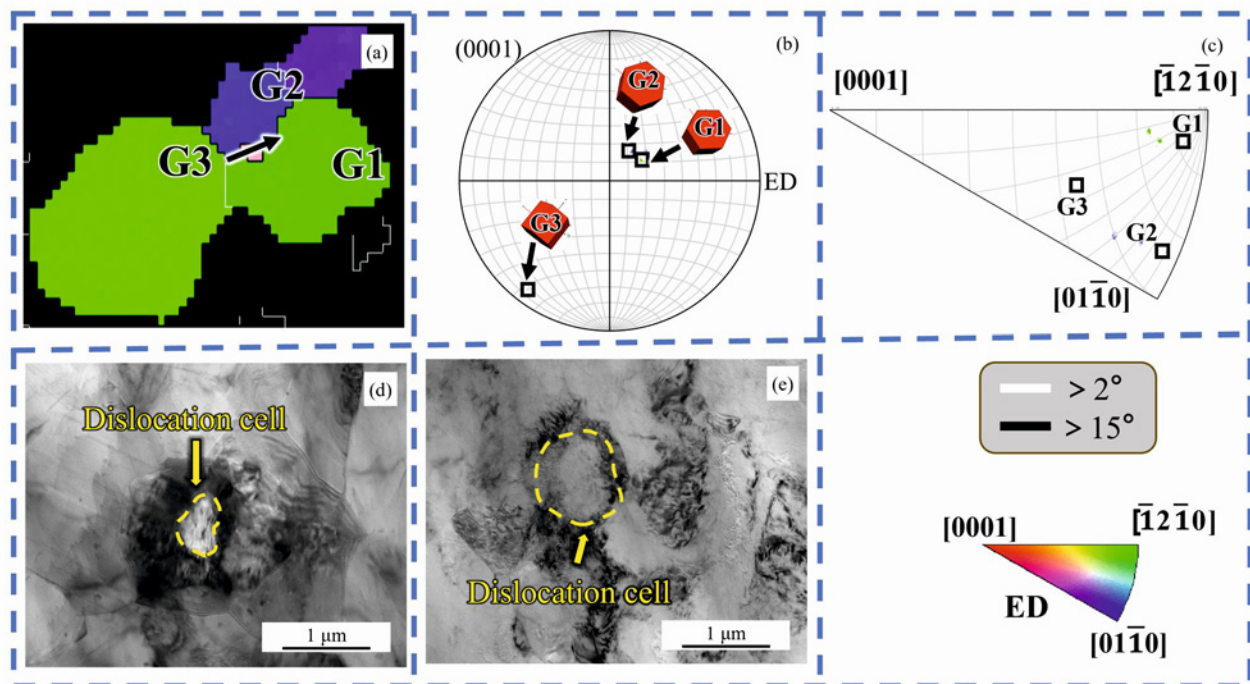


Fig. 14. EBSD analysis of region C in E3.0 alloy and TEM images: (a) IPF map, (b) (0001) pole figure and 3D orientation map, (c) inverse pole figure, (d) and (e) TEM images showing second phase particles and dislocation configurations.

the alloy's plasticity [38]. Additionally, nanoscale second phase particles dispersed in the matrix can pin grain boundaries, strengthen the matrix, and improve the alloy's mechanical strength. Thus, at 0–3.0 wt.% Y, increasing Y content promotes continuous grain refinement and a growing number of dispersed nanoscale second phase particles, leading to continuous improvements in the alloy's strength and plasticity.

At 4.0–5.0 wt.% Y, as evidenced in Figs. 1, 2, 6, and 7, increasing Y content continues to refine the alloy's grains while increasing the size and quantity of second phase particles. Notably, larger second phase particles exhibit obvious agglomeration in localised regions of the alloy, a phenomenon particularly pronounced in Mg alloys extruded with high Y contents. Such agglomeration markedly degrades microstructural uniformity. Grain refinement further enhances the fine-grain strengthening effect, which would otherwise contribute to increased strength and plasticity with rising Y content. During tensile deformation, dislocations preferentially accumulate near larger second phase particles ($> 1 \mu\text{m}$), inducing local stress concentration and crack initiation [39]. Additionally, larger second phase particles ($> 1 \mu\text{m}$) typically form a non-coherent interface with the alloy matrix, resulting in weak interfacial bonding between the particles and the matrix. During tensile deformation, these particles easily detach from the matrix, inducing early microvoid nucleation and reducing the alloy's plasticity [40]. Thus, with increasing Y content, while continuous grain refinement enhances the alloy's strength and

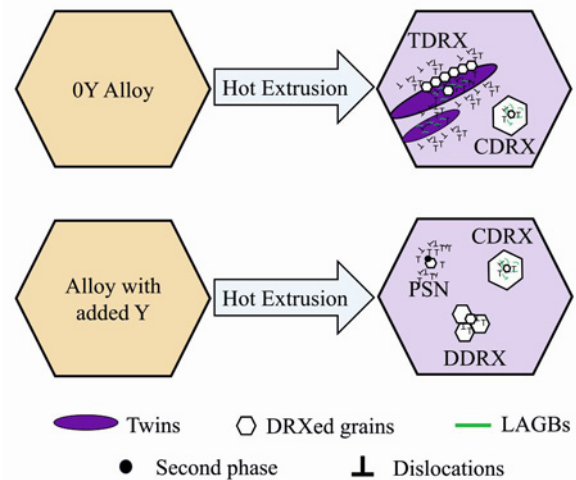


Fig. 15. Schematic illustration of DRX mechanisms in as-extruded $\text{Mg}_{2.0}\text{Zn}_{2.0}\text{Al}_{0.3}\text{Mn}_x\text{Y}$ alloys under different Y contents.

plasticity, the growth of second phase particles and expanded agglomeration regions exert a detrimental effect on plasticity. Consequently, the alloy exhibits a trend of increasing strength and decreasing plasticity.

4.3. Microstructure influence on corrosion behavior

Generally, the corrosion resistance of Mg alloys is

governed by composition, grain size, dislocation density, and the characteristics of second phase particles. In the E0 alloy, as alloying elements are fully dissolved in the matrix (Fig. 1), corrosion behavior is primarily influenced by grain size, twinning, and dislocation density.

The native MgO/Mg(OH)₂ film on Mg alloys is loose and porous, providing less protective capability than the passive film on Al alloys. Nevertheless, its initial formation and breakdown are still affected by the microstructure. Coarse grains can lead to non-uniform oxide film growth, increase susceptibility to localised breakdown by Cl[−], and facilitating pitting corrosion. Surface MgO film growth in Mg alloys occurs mainly via ion diffusion along grain boundaries. Larger grain sizes result in slower and less uniform film formation as well as reduced film stability. Twin regions contain higher-energy atoms, making them preferential sites for corrosion initiation [41]. High dislocation density causes lattice distortion, which in turn alters the surface potential distribution and gives rise to microgalvanic effects [42]. These factors collectively enhance the alloy's corrosion susceptibility. As shown in Fig. 11a, deep corrosion pits (−57 μm) formed in the Y-free alloy matrix following accelerated penetration of the corrosive medium along distorted lattice pathways. Thus, the combination of coarse grains, abundant tensile twins, and high dislocation density results in inferior corrosion resistance in the Y-free alloy.

In addition, crystallographic texture also influences the corrosion behavior of the alloy. It has been reported that the surface energies of Mg on the (0001), (10 $\bar{1}$ 0), and (11 $\bar{2}$ 0) planes are 1.808, 1.868, and 2.156 eV nm^{−2}, respectively [43]. The closely packed (0001) basal plane exhibits a lower dissolution rate compared to the prismatic and pyramidal planes. With increasing Y content, the fraction of grains oriented with <0001> parallel to the extrusion direction (ED) increases, which contributes to the gradual improvement in corrosion resistance. As evidenced in Figs. 11b–d, the corrosion morphology becomes more uniform with increasing Y content, indicating that continued grain refinement enhances the microstructural homogeneity of the alloy matrix and improves the protective quality of the surface oxide film. Thus, it can be inferred that although the increasing number of nanoscale second phase particles with higher Y content enhances the driving force for microgalvanic corrosion, the negative effect caused by these particles is outweighed by the protective effect provided by the denser, more continuous oxide film. Therefore, within the Y content range of 1.0–3.0 wt.%, the corrosion resistance of the alloy gradually improves due to the combined effects of grain refinement, enhanced compactness and continuity of the oxide film, and the increased fraction of grains with the <0001>//ED orientation.

At 4.0–5.0 wt.% Y, as evidenced in Figs. 12e,f, the number and size of corrosion pits aligned along with the ED increase with Y content. This phenomenon is attributed to microgalvanic corrosion induced by large second phase particles distributed along the ED. Although grain refinement enhances the compactness and continuity of the surface protective oxide film, the growing size and quantity of second phase particles increase the driving force for microgalvanic corrosion with the matrix. Additionally, the introduction of more reaction sites enhances the ability of chloride ions to corrode the protective oxide film. Thus, at 4.0–5.0 wt.% Y, increasing Y content leads to larger and more numerous second phase particles, which accelerates matrix dissolution and leads to a gradual decrease in the alloy's corrosion resistance.

5. Conclusions

This study systematically elucidates the critical role of Y content in tailoring the microstructure, mechanical properties, and corrosion behavior of extruded Mg-Zn-Al-Mn alloys. The primary significance of our findings lies in identifying a critical Y content window (approximately 3.0–4.0 wt.%) that enables an optimal synergy between strength and corrosion resistance, driven by distinct microstructural evolution mechanisms.

(1) The addition of Y fundamentally alters the recrystallisation mechanism. Without Y's participation, complete DRX is driven by TDRX, resulting in coarse grains. With 1.0–5.0 wt.% Y, the size and volume fraction of Al₂Y and (Al, Zn)₁₁Y₃ second phase particles increase significantly. These particles promote grain refinement via PSN and exert a strong Smith-Zener pinning effect, thereby simultaneously enhancing basal texture strength and progressively suppressing twin formation.

(2) A critical Y content window (3.0–4.0 wt.%) for optimising mechanical properties is identified. Below this range (0–3.0 wt.%), continuous grain refinement and the dispersion of nanoscale precipitates lead to simultaneous improvements in strength and plasticity via the Hall-Petch and Orowan strengthening mechanisms. However, above 4.0 wt.%, although grain refinement continues to enhance strength, the coarsening and aggregation of second phase particles act as stress concentration sites, causing a gradual deterioration in plasticity.

(3) The effect of Y on corrosion resistance is dualistic and microstructure dependent. At lower contents (0–3.0 wt.%), progressive grain refinement improves the compactness and protectiveness of the surface film, leading to more uniform corrosion and enhanced resistance. Conversely, at higher contents (4.0–5.0 wt.%), the increased population of coarse second

phase particles amplifies the microgalvanic corrosion effect, accelerating matrix dissolution and shifting the corrosion mode towards localised attack.

The key significance of this work is the established principle that the optimal alloy performance is achieved not by maximising grain refinement alone, but by precisely controlling the size and distribution of second phases through Y alloying. This insight provides a crucial guideline for designing high-performance Mg-Zn-Al-Mn-Y alloys, emphasising the importance of balancing grain refinement with second phase engineering to avoid detrimental microgalvanic effects while leveraging strengthening mechanisms. Therefore, this study is hoped to offer a coherent strategy for developing Mg alloys with a superior property profile for demanding structural applications.

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