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Research Article

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Posted Date: July 23rd, 2026

DOI: <https://doi.org/10.21203/rs.3.rs-10087563/v1>

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Additional Declarations: No competing interests reported.

Version of Record: A version of this preprint was published at International Journal of Metalcasting on October 5th, 2026. See the published version at <https://doi.org/10.1007/s40962-026-02148-2>.

Effect of Mg-Controlled Solidification Pathways on Microstructure and Properties of As-Cast Al-1Y-xMg Alloys

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Abstract

Understanding the relationships among solidification processing, microstructure, and properties is central to the design of advanced aluminum alloys. In this work, as-cast Al-1Y-xMg ($x = 0-8$ wt.%) alloys were prepared via conventional casting to investigate how Mg content controls solidification pathways and, consequently, the resulting microstructure and associated mechanical and thermal properties. With increasing Mg content, the solidification pathway evolves from isolated Y-rich phases to sequential precipitation of τ -Al_xMgY and β -Al₃Mg₂, ultimately forming a continuous $\tau+\beta$ eutectic network at high Mg levels. The alloy with 6 wt.% Mg achieves a peak tensile strength of 158.0 MPa—approximately 95% higher than the Mg-free alloy—while retaining 11.0% uniform elongation, demonstrating a favorable strength-ductility balance. Concurrently, the eutectic network topology governs fracture behavior, transitioning from ductile dimple to quasi-cleavage, and mediates dynamic strain aging via Portevin-Le Chatelier serrations. Physically, electrical and thermal conductivities are continuously tunable within 28.2-51.8% IACS and 148-218 W·m⁻¹·K⁻¹, respectively. The dominant electron scattering mechanism shifts from solute-controlled scattering at low Mg contents to interface-controlled scattering at high Mg contents, directly correlating with the evolving eutectic network. This study establishes quantitative processing-microstructure-property relationships in Al-Mg-Y alloys, offering guidance for tailoring cast aluminum alloys for combined structural and thermal management applications.

Keywords

Al-Mg-Y alloys; Solidification pathway; Eutectic network; Mechanical properties; Electrical conductivity

1 Introduction

Al-Mg alloys are widely employed in lightweight structures for automotive, aerospace, and marine applications due to their excellent corrosion resistance, good formability, and weldability [1-4]. Their primary strengthening mechanism is solid-solution hardening by Mg atoms, which generate significant lattice distortions owing to an approximately +11.8% atomic radius mismatch with the Al matrix, effectively impeding dislocation motion [5, 6]. Commercial Al-Mg alloys typically contain 1-10 wt.% Mg, and increasing Mg content generally enhances strength [7, 8]. However, high Mg concentrations introduce a fundamental challenge: although the equilibrium solubility of Mg in Al can reach 17.4 wt.% at the eutectic temperature of 451 °C, conventional casting and solution treatment frequently lead to inhomogeneous precipitation of the β -Al₃Mg₂ phase along grain boundaries [9]. This brittle body-centered cubic intermetallic forms incoherent interfaces with the matrix, promoting stress concentration and microvoid nucleation during deformation, which facilitates crack propagation and drastically reduces ductility [10]. Moreover, a continuous network of β phase along grain boundaries exacerbates strain-hardening heterogeneity [11], diminishes the softening resistance of weld heat-affected zones [12], and acts as a cathodic site in chloride-containing environments, markedly increasing susceptibility to stress corrosion cracking [13]. This inherent trade-off between strength and ductility, combined with rising demands for functional properties such as thermal and electrical conductivity, poses a critical bottleneck for the application of medium-to-high Mg Al-Mg alloys in load-bearing and heat-dissipating components [14].

To address these challenges, microalloying strategies have been extensively explored to manipulate the microstructure. Among them, scandium (Sc) addition has proven highly effective in enhancing overall alloy performance [15-17]. For instance, Zou et al. prepared Al-Ti-Sc master alloys via aluminothermic reduction, significantly mitigating the brittleness of high-Mg alloys [18]. Wang et al. reported that primary Al₃(Sc, Zr) phases formed in the melt exhibit very low lattice mismatch with α -Al, acting as efficient heterogeneous nucleation sites that refine cast grains into uniform equiaxed structures [19]. Despite these advantages, Sc-containing alloys face two major limitations: Sc is prohibitively expensive, several hundred times the cost of pure Al [20], and the precipitation behavior of Al₃Sc dispersoids is highly sensitive to cooling rate, making it difficult to achieve optimal dispersoid distributions in thick industrial castings, thereby compromising strengthening effects [21, 22]. More cost-effective rare earth elements, such as cerium (Ce), erbium (Er), and yttrium (Y), have therefore received increasing attention. Jin et al. showed that adding Ce to Al-5Mg alloys significantly refines α -Al grains, improves microstructural homogeneity, and enhances hot-crack resistance [23]. However, the equilibrium solubility of Ce in Al is nearly zero, and

it mainly exists as rod-like $\text{Al}_{11}\text{Ce}_3$ phases at grain boundaries, contributing little to high-temperature strength; excessive addition can produce coarse needle-like phases that fracture the matrix and reduce ductility [24]. Er has been used to form coherent $\text{L}_{12}\text{-Al}_3\text{Er}$ nanoprecipitates, improving thermal stability [25], but its maximum solubility in Al is extremely low (<0.05 wt.%), limiting the precipitate volume fraction and the theoretical strengthening increment [26]. In contrast, Y can form multiple phases in Al, such as Al_3Y and Al_4MgY , offering the potential to simultaneously enhance strength and thermal stability [27, 28]. The phase equilibria in the Al-Mg-Y system are more complex, and adjusting the Mg/Y ratio provides a promising route for tailoring phase composition and morphology. Nonetheless, the underlying mechanisms, particularly under non-equilibrium solidification, remain poorly understood. Although yttrium can provide multiphase strengthening in Al-Mg alloys, the specific manner in which its effects evolve with Mg content remains unclear. Existing studies have primarily focused on either very low or very high Y contents, whereas systematic investigations that fix the Y content at 1 wt.% to examine the influence of Mg on solidification pathways and microstructural evolution remain scarce. As a result, key aspects of Mg content dependent nonequilibrium solidification including phase selection sequences, eutectic network formation, and the resulting structure-property relationships have not yet been systematically clarified.

To address these gaps, a series of as-cast Al-1Y-xMg ($x = 0\text{-}8$ wt.%) alloys were designed and fabricated. Multiscale characterization was conducted to clarify the role of Mg content in governing solidification behavior. The results reveal a continuous microstructural evolution from isolated Y-rich phases to a fully developed $\tau+\beta$ eutectic network, accompanied by a well-defined intermetallic precipitation sequence and morphological transition. Based on this structural evolution, the correlations between eutectic network topology and mechanical response, including strength, ductility, and fracture behavior, are quantitatively established. The influence of Mg content on dynamic strain aging and its role in ductility regulation is further clarified. In addition, the coupled effects of Mg solutes and grain boundary secondary phases on electron and phonon scattering are systematically analyzed. Collectively, this work provides a quantitative framework for understanding and tailoring the coupled structural and functional properties of Al-Mg-Y alloys.

2 Experimental Procedure

The raw materials used in the experiment were Al-20Mg master alloy, Al-10Y master alloy, and pure aluminum. The casting experiments were conducted using a crucible resistance furnace to prepare Al-1Y-xMg alloys ($x = 0, 2, 4, 6, 8$ wt.%), each with a total mass of 500 g. The burn-off loss for Mg was approximately 6%, and for Al, it was about 4%, while the losses of other alloying elements were negligible.

The actual compositions of these alloys are shown in Table 1. Although deviations exist between the actual and nominal compositions, the nominal compositions are used throughout this paper for ease of discussion. The samples are designated as 0Mg, 2Mg, 4Mg, 6Mg and 8Mg, with their composition values located within the red stars marked on the phase diagram (Fig. 1).

Table 1 The actual and nominal compositions of the designed Al-1Y-xMg alloys.

Alloy	Nominal compositions (wt%)			Actual compositions (wt%)					
	Al	Mg	Y	Al	Mg	Y	Mn	Fe	Zn
Al-1Y	99	0	1	98.42	0.01	1.19	0.08	0.28	0.02
Al-2Mg-1Y	97	2	1	96.57	1.94	1.13	0.09	0.25	0.02
Al-4Mg-1Y	95	4	1	94.77	3.86	1.07	0.06	0.21	0.03
Al-6Mg-1Y	93	6	1	92.64	5.87	1.12	0.07	0.28	0.02
Al-8Mg-1Y	91	8	1	90.63	7.92	1.10	0.11	0.22	0.02

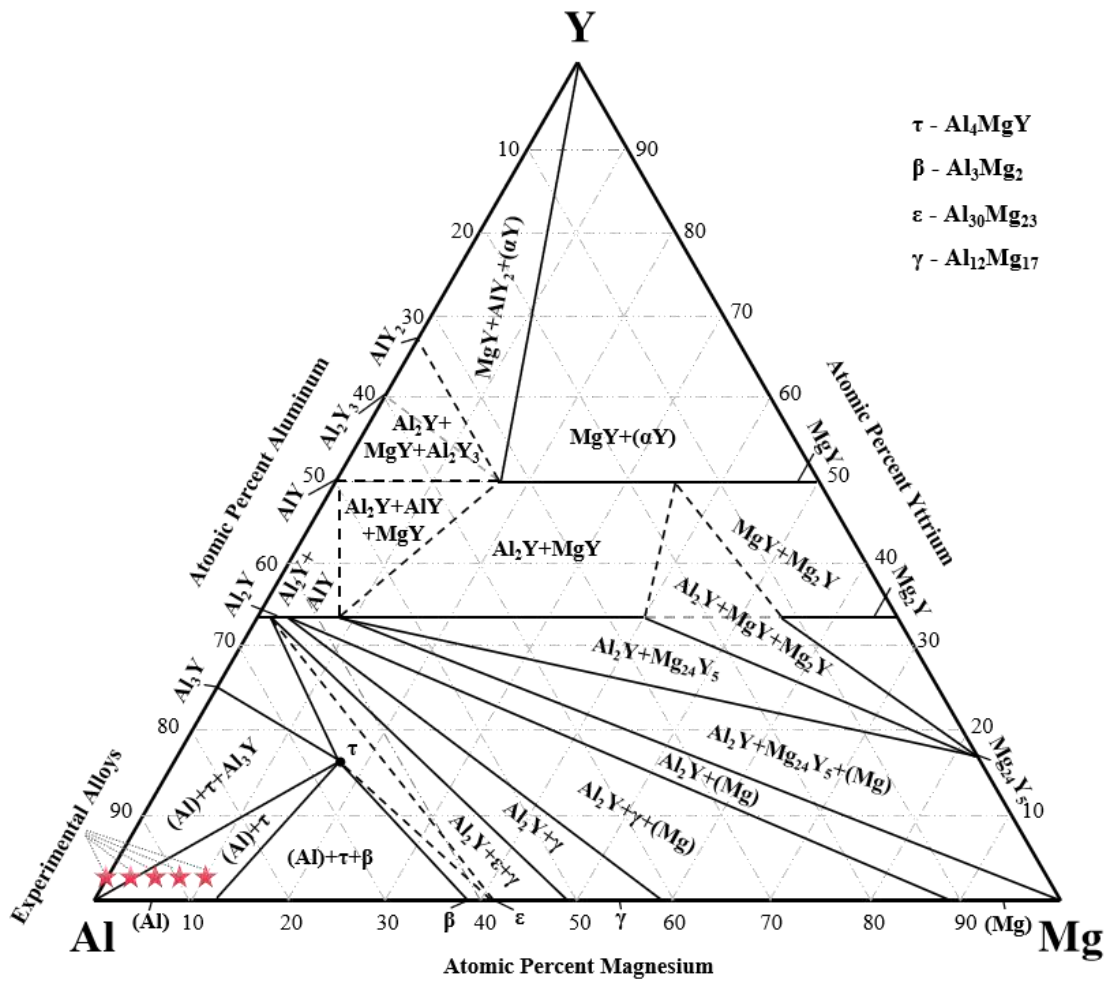


Fig. 1. Isothermal section of the Al-Mg-Y system at 400 °C [29-31].

Prior to the experiment, the raw materials and tools were placed in a muffle furnace at 200 ± 5 °C for approximately one hour for preheating. The preheated pure aluminum and master alloys were then placed

into a graphite crucible and loaded into the resistance furnace, with the temperature set to 800 ± 30 °C. After the alloys were completely melted, the melt was manually stirred at approximately 30 rpm for 1 minute to promote compositional homogeneity. Subsequently, a covering flux (1:1 mass ratio mixture of NaCl and KCl), used to prevent air ingress into the melt during pouring, was added, and the melt was held for 10 minutes before skimming off the dross. The first refining step was then conducted by adding a refining agent (C_2Cl_6) and stirring. After the chlorine gas had substantially volatilized, the melt was held for 15 minutes, followed by dross removal. The second refining step repeated this refining procedure. The covering flux was added again in preparation for tapping. After holding for 5 minutes, the melt was tapped and allowed to cool. The temperature was monitored using a thermocouple, and once it reached 650 ± 10 °C, the melt was poured into a mold and air-cooled to room temperature. Finally, the cast bars were machined into metallographic samples and tensile test specimens. The preparation flowchart is shown in Fig. 2.

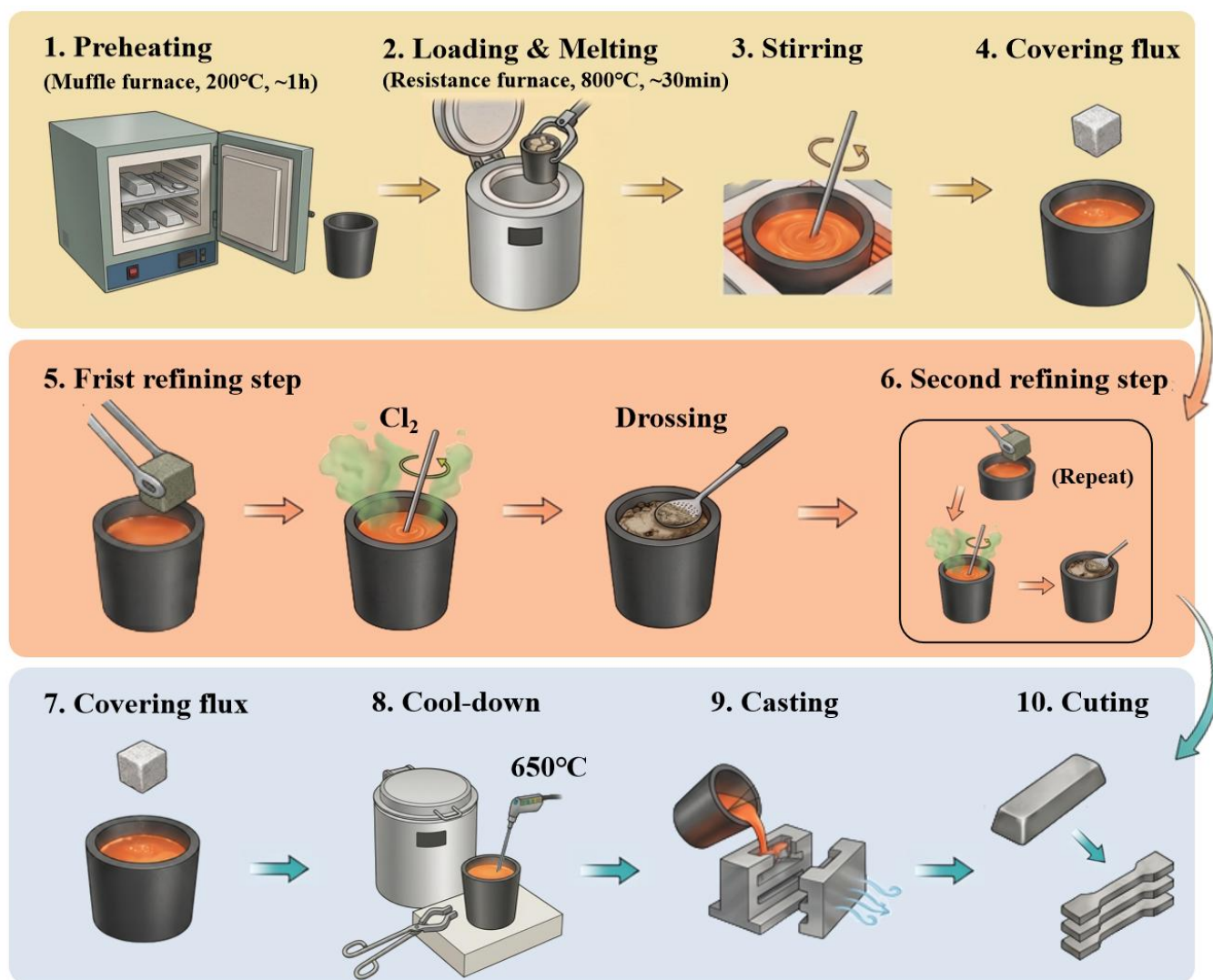


Fig. 2. Schematic diagram of the experimental process

The alloy melting and casting were performed using a Shanghai Hua Yan SG2-7.5-10 well-type resistance furnace. Samples were sectioned using a Denmark Struers Labotom-15 manual metal cutter,

starting 20 mm from the bottom of each ingot. These samples were then ground and polished in water until a mirror-like, scratch-free surface was achieved, followed by etching with Keller's reagent for 20 seconds. The microstructure was observed and imaged using an optical microscope (OM, Zeiss Observer. A1, Germany). Microstructural characterization was performed using a TM400PLUS benchtop scanning electron microscope (SEM), an IXRF5500 energy dispersive spectrometer (EDS). Each cast ingot was cut into three tensile specimens. Tensile tests were conducted using an Instron 8801 universal testing machine at a constant strain rate of 0.5 mm/min. The fracture surfaces of the tested specimens were examined by SEM to analyze morphological changes. Qualitative phase analysis of the experimental samples was carried out using a Rigaku D/MAX 2500 V X-ray diffractometer (XRD). Vickers microhardness was measured under a 1000 gf load (HV-1000SPTA), averaging five indentations per sample. Rectangular bars (20 mm × 4 mm × 4 mm) were ground, cleaned, and dried; electrical conductivity was measured using a Sigma 2008A eddy current meter and converted to thermal conductivity. Differential scanning calorimetry (DSC) was performed on an STA 449 F3 Jupiter at 10 °C/min from room temperature to 800 °C then cooled to 400 °C. Open-circuit potential (OCP) tests were carried out in a three-electrode cell (CHI600F workstation) with the as-cast alloy as working electrode (14 mm diameter), a platinum plate as counter electrode, and a saturated calomel electrode (SCE) as reference. Specimens were immersed in 3.5 wt.% NaCl for 30 min to stabilize the surface, followed by 10 min OCP recording.

3. Results

3.1 Microstructural Evolution

Fig. 3 systematically illustrates the microstructural evolution of Al-1Y-xMg ($x = 0, 2, 4, 6, 8$ wt.%) alloys in the as-cast condition. In the absence of Mg addition (Fig. 3(a)), the matrix consists of equiaxed α -Al grains with clearly defined, bright white grain boundaries, where Y is predominantly segregated. When the Mg content increases to 2 wt.% (Fig. 3(b)), the overall microstructure remains largely unchanged, with only slight variations in contrast among different regions attributable to compositional differences. At 4 wt.% Mg (Fig. 3(c, d)), blocky Y-rich phases are first observed along the grain boundaries, reaching sizes of several tens of micrometers and exhibiting sharp interfaces with the α -Al matrix. With Mg content elevated to 6 wt.% (Fig. 3(e, f)), the continuity of grain boundaries begins to decrease, and dark granular β -Al₃Mg₂ phases precipitate along the boundaries, coexisting adjacent to the bright Y-rich phases. Upon further increasing Mg to 8 wt.% (Fig. 3(g, h)), the amount of dark β -phase increases, forming clusters or continuous distributions, leading to further degradation of grain boundary continuity. Additionally, newly formed brownish τ -Al_xMgY phases are observed interspersed within the microstructure.

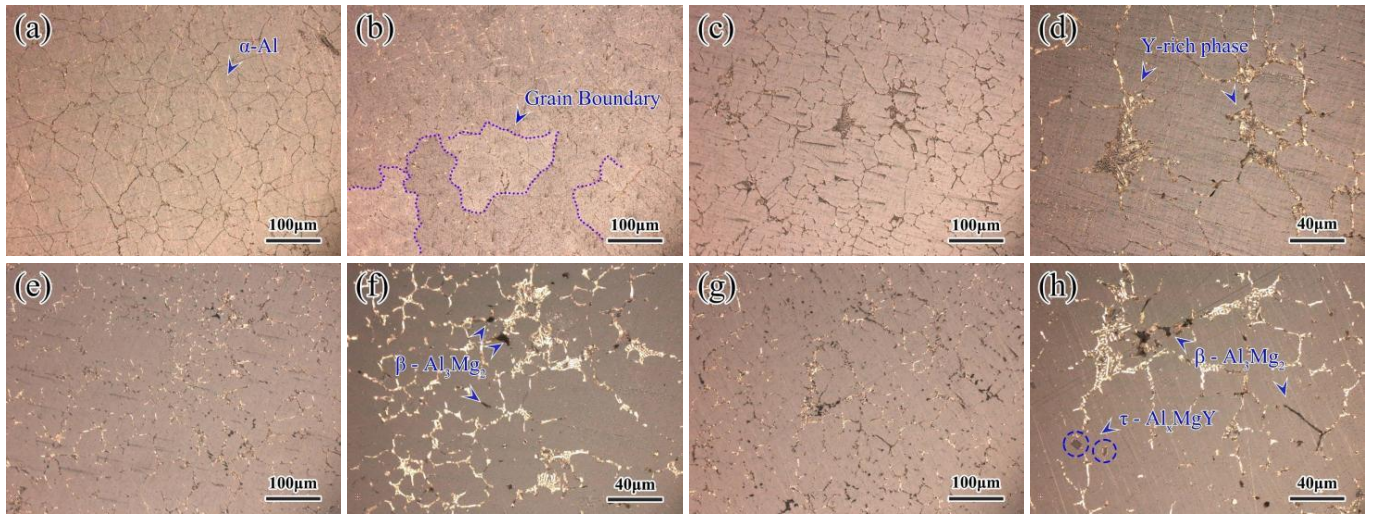


Fig. 3. Optical micrographs of as-cast Al-1Y-xMg alloys: (a) 0Mg, (b) 2Mg, (c, d) 4Mg, (e, f) 6Mg, and (g, h) 8Mg.

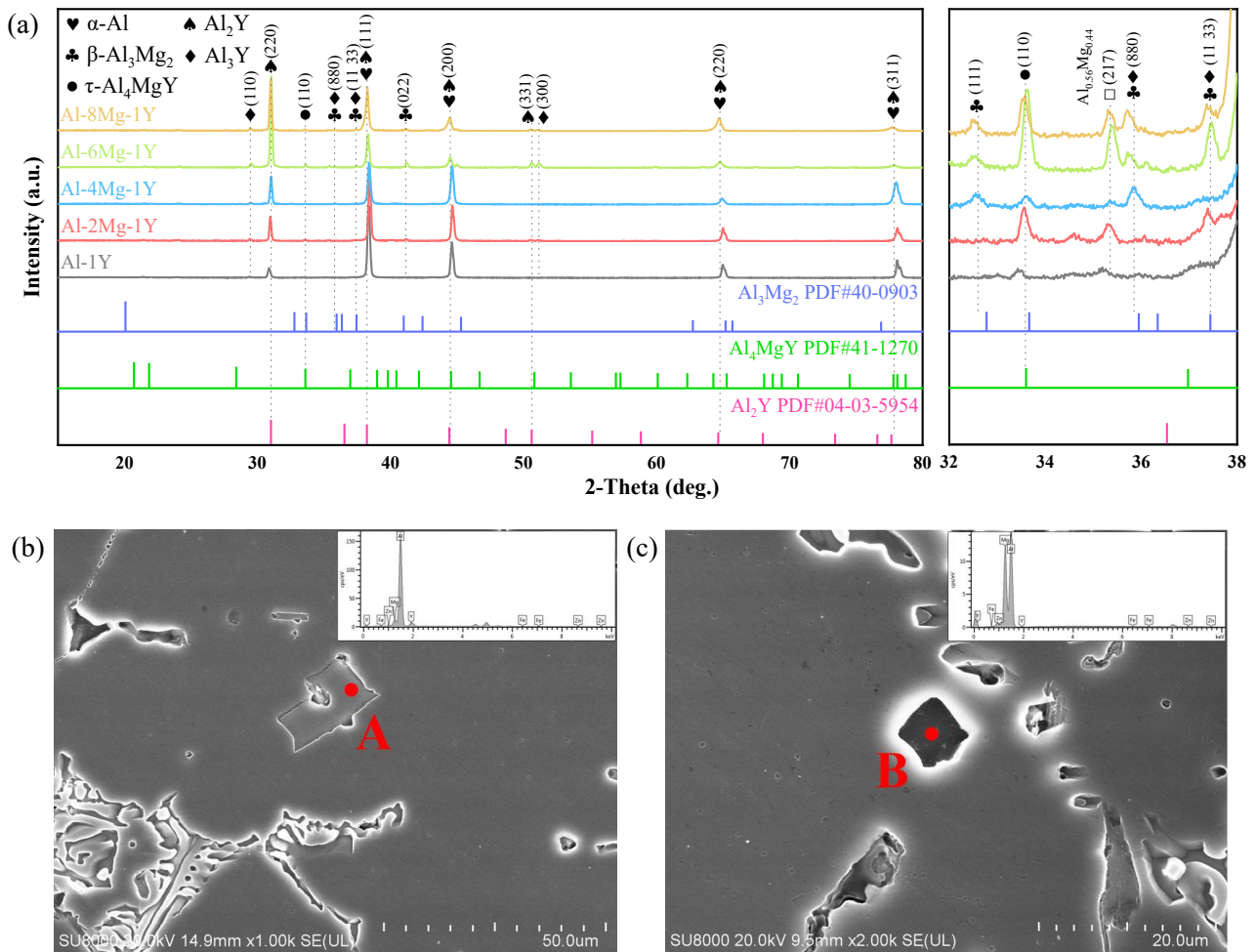


Fig. 4. (a) XRD patterns of the Al-1Y-xMg alloys with varying Mg contents; (b) EDS mapping of selected precipitated secondary phases in the alloy.

Fig. 4(a) shows the XRD patterns and quantitative analysis of the Al-1Y-xMg alloys with varying Mg contents. All samples exhibit sharp diffraction peaks at $2\theta = 38.5^\circ$, 44.7° , 65.1° , and 78.2° , corresponding to the (111), (200), (220), and (311) planes of face-centered cubic (fcc) α -Al, respectively. Within the 2θ

range of 32°-38°, diffraction peaks of β -Al₃Mg₂ and τ -Al_xMgY emerge with increasing Mg content, consistent with the metallographic observations. The EDS mappings and point scan results of these secondary phases are presented in Fig. 4(b, c) and Table 3. Table 2 summarizes the XRD line broadening analysis results for the Al-1Y-xMg alloys. Using a combined approach of the Scherrer equation and the Williamson-Hall (W-H) method, after subtracting instrumental broadening, the crystallite size D_{XRD} was calculated from the α -Al (111) peak, and the microstrain ϵ was derived from the slope of the W-H linear fit [32]. The dislocation density ρ was calculated using Eq. (1):

$$\rho = 2\sqrt{3} \frac{(\epsilon^2)^{\frac{1}{2}}}{bD_{\text{XRD}}} \quad (1)$$

where $b = 0.286$ nm represents the magnitude of the Burgers vector for Al. With increasing Mg content, D_{XRD} fluctuates within the range of 78-103 nm, reaching a maximum at 2 wt.% Mg (103 nm). Correspondingly, the lattice strain ϵ increases from 1.09×10^{-3} to 1.43×10^{-3} and then decreases to 1.32×10^{-3} , while the dislocation density ρ rises from $0.65 \times 10^{14} \text{ m}^{-2}$ to $1.12 \times 10^{14} \text{ m}^{-2}$ before declining to $0.93 \times 10^{14} \text{ m}^{-2}$. The results indicate that a trace amount of Mg (2 wt.%) promotes subgrain coalescence, leading to the lowest dislocation density. With further Mg addition, the dislocation density increases again due to enhanced solute stress and precipitation, yet it remains on the order of 10^{14} m^{-2} , and the alloy retains a nano/submicrocrystalline structure.

Table 2 FWHM analysis from the α -Al (111) peak in Al-1Y-xMg alloys with different Mg contents.

Alloy	FWHM (°)	D_{XRD} (nm)	$\langle \epsilon^2 \rangle^{1/2}$ (%)	$P (10^{14} \text{m}^{-2})$
Al-1Y	0.231	89	1.26	0.87
Al-2Mg-1Y	0.205	103	1.09	0.65
Al-4Mg-1Y	0.245	81	1.39	1.04
Al-6Mg-1Y	0.251	78	1.43	1.12
Al-8Mg-1Y	0.236	85	1.32	0.93

Table 3 EDS point analysis of selected secondary phases in the alloy.

Points (at%)	Al	Mg	Y	Zn	Fe	Phase
A	89.44	5.16	5.33	0.00	0.06	τ -phase
B	66.14	33.40	0.02	0.35	0.09	β -phase

3.2 Mechanical Properties and Fracture Morphology

Fig. 5(a) presents the engineering stress-strain curves of as-cast Al-1Y-xMg alloys at room temperature, with the corresponding values summarized in Table 4. Due to the continuous nonlinear transition of the curves and the presence of serrated flow characteristic of dynamic strain aging (DSA) throughout the

deformation, the materials exhibit no distinct yield point [33]. Therefore, the ultimate tensile strength (UTS) and elongation (EL) are adopted in this study as metrics for strength and ductility, respectively. With increasing Mg content, the strength of the alloys initially increases and then decreases: the UTS increases monotonically from 81.2 MPa at 0 wt.% Mg to 158.0 MPa at 6 wt.% Mg, but decreases to 151.5 MPa at 8 wt.% Mg. Meanwhile, the uniform elongation consistently declines from 25.6% to 8.0%. In the range of 4–6 wt.% Mg, the strength gain slows while ductility continues to drop markedly, indicating diminishing marginal benefits with further Mg additions. The peak to valley stress amplitude of the serrations increases with Mg content, reaching (6.0 ± 0.5) MPa in the 6 wt.% Mg alloy, accompanied by an increased serration frequency. This reflects enhanced solute atom-dislocation interactions, which, together with the continuous nonlinear stress-strain response, account for the absence of macroscopic yielding.

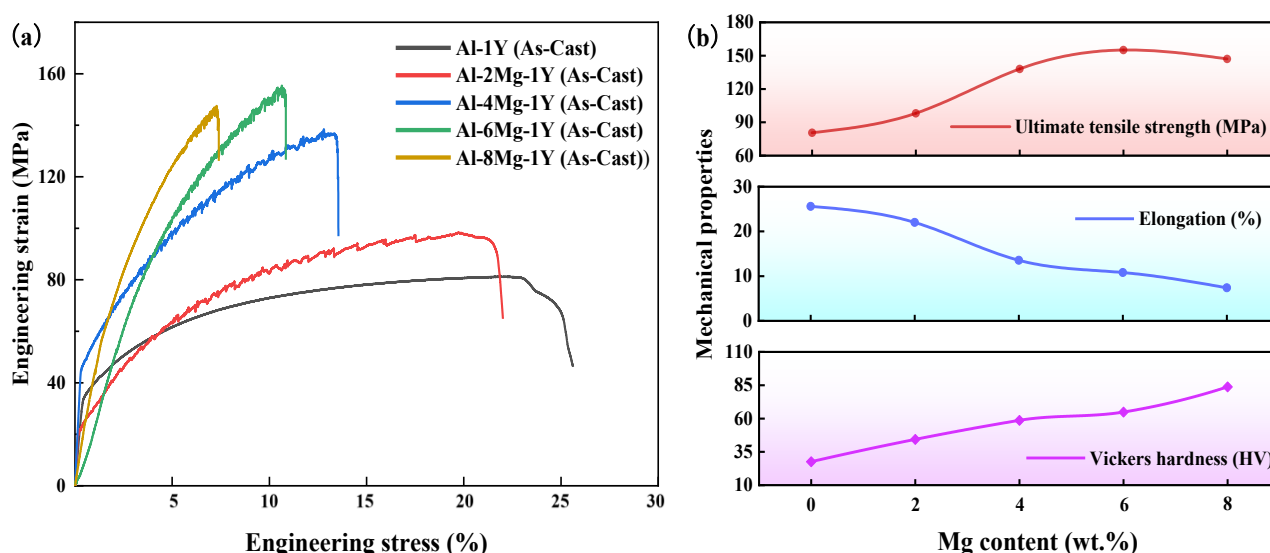


Fig. 5. (a) Mechanical properties of as-cast Al-1Y-xMg alloys (b) UTS, elongation and Vickers hardness

Table 4 Mechanical properties of Al-1Y-xMg alloys at different Mg contents.

Alloy	UTS (MPa)	EL (%)	HV
Al-1Y	81±3	25.6±2.4	27.6±0.6
Al-2Mg-1Y	91±4	22.4±2.0	44.3±0.5
Al-4Mg-1Y	138±7	13.0±1.7	58.6±1.1
Al-6Mg-1Y	158±8	11.1±1.8	64.8±0.1
Al-8Mg-1Y	151±8	8.0±2.0	83.8±1.1

Fig. 5(b) illustrates the variation in mechanical properties of as-cast Al-1Y-xMg alloys with Mg content. As shown, the Vickers hardness (HV) increases monotonically with rising Mg content: from (27.6 ± 0.7) HV at 0 wt.% Mg to (83.8 ± 1.1) HV at 8 wt.% Mg, representing an increase of approximately 200%. This trend is consistent with the tensile strength peak observed in the 0–6 wt.% Mg range. At 8 wt.% Mg,

although the hardness continues to rise due to extensive precipitation of the β -Al₃Mg₂ phase, the ultimate tensile strength shows a slight decline. Moreover, a strong negative linear correlation ($R^2 \approx 0.97$, Fig. 6) is observed between HV and EL, indicating that every increase of approximately 10 HV corresponds to a decrease in EL of about 2.5%. Therefore, microhardness can be used as a rapid indicator to estimate both the strength level and ductility reserve of as-cast Al-Mg-Y alloys, confirming the dominant role of solid solution strengthening in the overall deformation resistance.

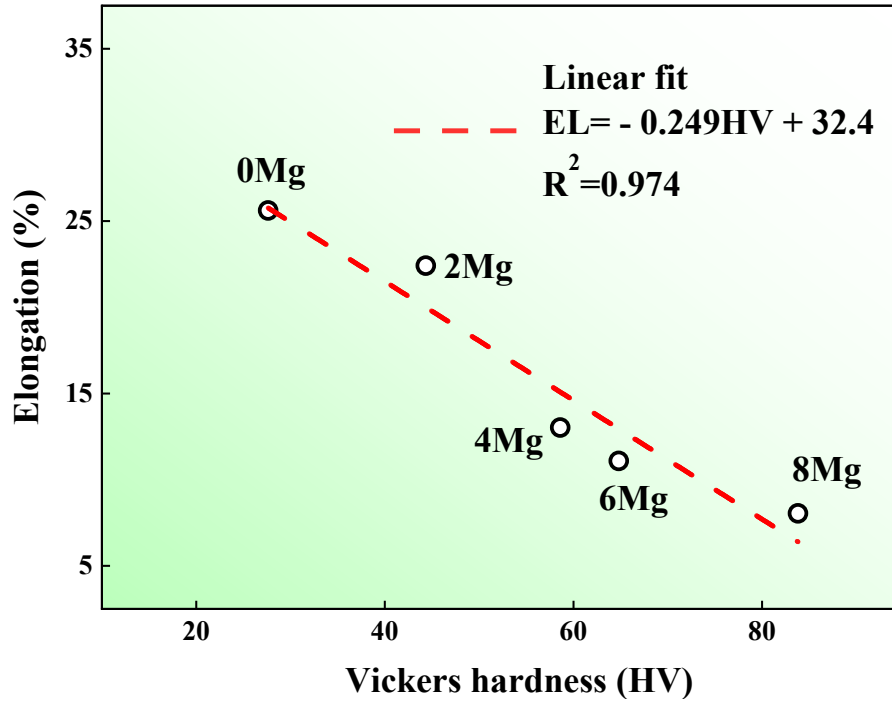


Fig. 6. Linear relationship between Vickers hardness and elongation in as-cast Al-1Y-xMg alloys ($EL = -0.249 HV + 32.4$, $R^2 = 0.974$).

Fig. 7 shows the fractography of tensile-fractured as-cast Al-1Y-xMg alloys, Fig. 7(a-e) correspond to 0-8 wt.% Mg, Fig. 7f is a magnified view of the boxed region in Fig. 7c. The low-Mg fractures (Fig. 7(a, b)) are dominated by equiaxed dimples with diameters of approximately 5-8 μm and scarce tear ridges, indicating sufficient plastic deformation prior to fracture. As the Mg content increases to 4 wt.% (Fig. 7c), the dimple size decreases to 2-4 μm with an inhomogeneous distribution, and short tear ridges appear locally, reflecting the suppression of dislocation emission at dimple bottoms by solid solution strengthening. The fracture surface of the 6 wt.% Mg alloy (Fig. 7d) exhibits mixed dimple-cleavage characteristics, with narrow strips of dimples connecting cleavage platforms, consistent with the reduced ductility reflected by the macroscopic EL of 11%. In the 8 wt.% Mg alloy (Fig. 7e), the area fraction of cleavage facets further increases, with the dimple region accounting for <30%, and tear ridges propagate linearly. The fracture mechanism transitions to quasi-cleavage dominated by cleavage, corresponding to the lowest EL (8%). Overall observations confirm that increasing Mg content shifts the fracture morphology from ductile

dimpling to brittle cleavage. This change in microscopic mechanism aligns with the monotonic decline in macroscopic ductility, further validating the weakening effect of solid solution strengthening on plasticity.

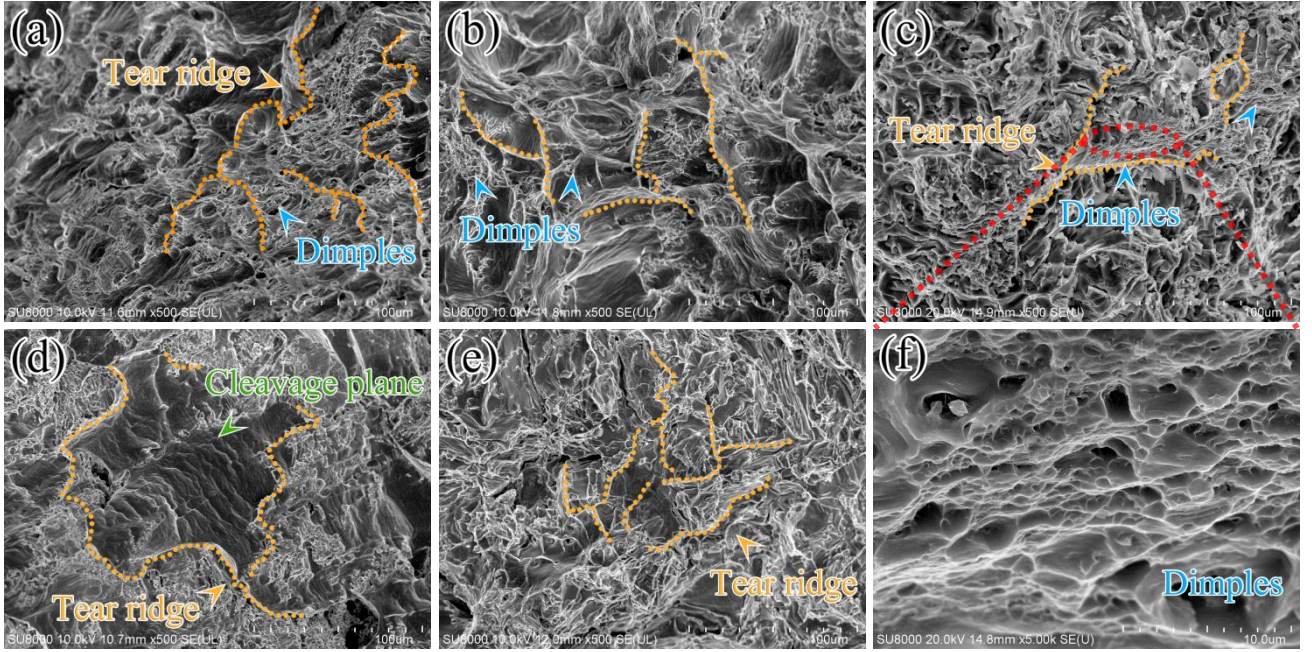


Fig. 7. SEM morphologies of the tensile fracture surfaces of as-cast Al-1Y-xMg alloys: (a) 0Mg, (b) 2Mg, (c) 4Mg, (d) 6Mg, (e) 8Mg, (f) higher-magnification view of the boxed region in Fig. 8(c).

3.3 Physical Properties

Fig. 8(a) shows the measured electrical conductivity (EC) of as-cast Al-1Y-xMg alloys at room temperature. The Wiedemann-Franz Law states that for many metallic materials at specific temperatures, the ratio of thermal conductivity (λ) to electrical conductivity (σ) is approximately constant [34]. Therefore, the thermal conductivity is calculated using Eq. (2). The formula is as follows:

$$k_e = L\sigma T \quad (2)$$

where $T = 293$ K and the Lorenz number $L = 2.45 \times 10^{-8} \text{ W} \cdot \Omega \cdot \text{K}^{-2}$. The calculated thermal conductivity exhibits an identical trend to that of the electrical conductivity: the EC decreases monotonically from 51.8 %IACS for Al-0Mg to 28.2 %IACS for Al-8Mg, corresponding to a decline in λ from $218 \text{ W} \cdot \text{m}^{-1} \cdot \text{K}^{-1}$ to $148 \text{ W} \cdot \text{m}^{-1} \cdot \text{K}^{-1}$. The absence of a recovery plateau across the entire composition range indicates that the increasing concentration of Mg in solid solution continuously enhances the scattering of electron and phonon transport. Consequently, electrical conductivity can serve as an effective indicator for rapidly assessing the thermal management capability of this alloy series.

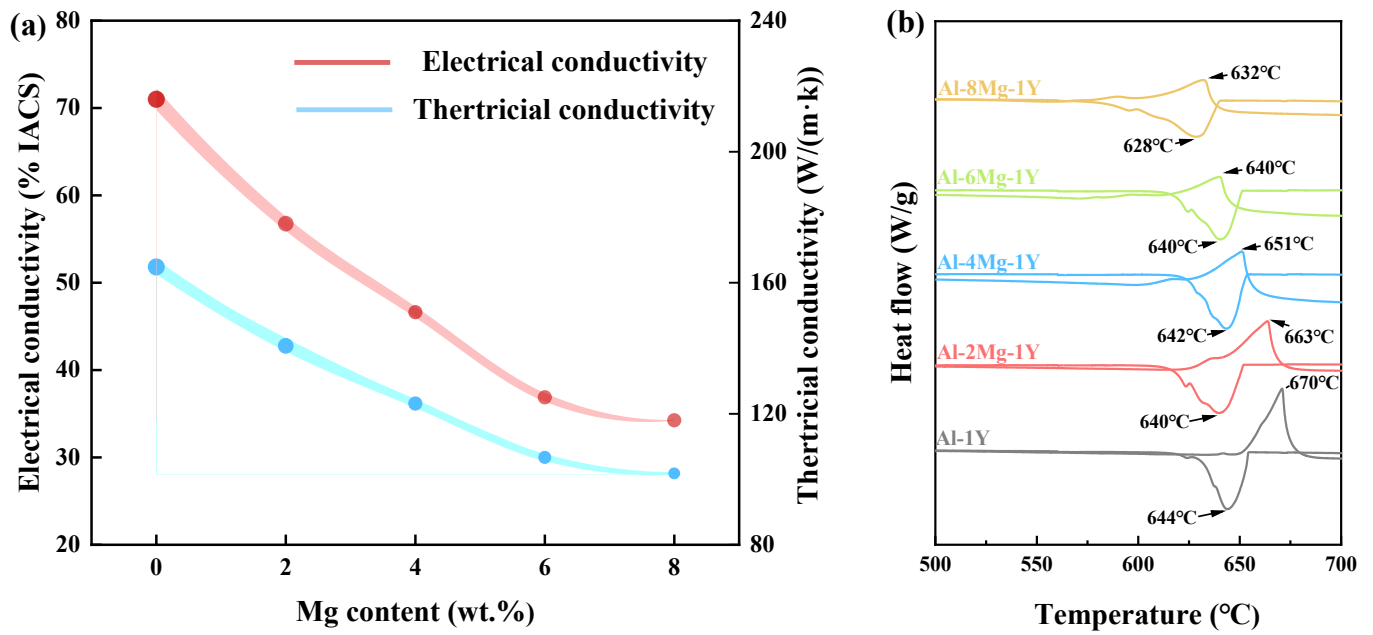


Fig. 8. (a) TC and EC of as-cast Al-1Y-xMg alloys, with the banded sections representing the margin of error, (b) DSC analysis of Al-1Y-xMg alloy with different Mg additions.

Table 5 Temperatures and enthalpy changes of Al-1Y-xMg alloys at different Mg contents.

Alloy	Heating 500-700°C				Cooling 700-500°C			
	T _{start} (°C)	T _p (°C)	T _{end} (°C)	ΔH (J·g ⁻¹)	T _{start} (°C)	T _p (°C)	T _{end} (°C)	ΔH (J·g ⁻¹)
Al-1Y	651	670	680	-205±1.8	654	644	626	202±2.0
Al-2Mg-1Y	640	663	671	-197±1.7	652	640	616	194±1.9
Al-4Mg-1Y	629	651	662	-189±1.9	653	642	620	186±2.1
Al-6Mg-1Y	617	640	648	-180±1.6	651	640	617	178±1.8
Al-8Mg-1Y	606	632	640	-172±2.0	640	628	597	170±2.2

Fig. 8(b) presents the Differential Scanning Calorimetry (DSC) heating curves of as-cast Al-1Y-xMg alloys at a constant heating rate of 10 °C·min⁻¹. All specimens exhibit a single endothermic peak during heating, corresponding to the melting of the α -Al matrix, with the peak temperatures (T_p) summarized in Table 5. As the Mg content increases, T_p gradually decreases from approximately 670 °C for the 0Mg alloy to about 632 °C for the 8Mg alloy, a reduction of nearly 38 °C. Concurrently, the melting enthalpy (ΔH) decreases monotonically from about 205 J·g⁻¹ for 0Mg to roughly 172 J·g⁻¹ for 8Mg, representing a decline of approximately 16%. These changes indicate that the progressive dissolution of Mg significantly weakens the bonding strength of the Al matrix, thereby depressing the melting point and suppressing the latent heat of fusion. Notably, for Mg contents $\geq$ 6 wt.%, the DSC curves display a broad shoulder on the low-temperature side of the main peak, attributable to the eutectic melting of Mg-rich phases (β -Al₃Mg₂ and τ -Al_xMg_y), with an onset temperature of about 520 °C. This confirms the formation of a eutectic network by

the continuously distributed β/τ phases during the final stages of solidification, as observed in the microstructure. No distinct exothermic peaks associated with recrystallization or solid-state phase transformations are detected during the cooling segment, suggesting the alloys remain in a metastable state after solidification without significant subsequent solid-state transitions. Overall, the DSC results quantitatively demonstrate that each 1 wt.% increase in Mg content lowers the solidus temperature by an average of approximately 4.7 °C. Furthermore, when $\text{Mg} \geq 6$ wt.%, the low-temperature eutectic reaction becomes a critical factor governing hot tearing susceptibility.

Open-circuit potential (OCP) measurements were carried out in a 3.5 wt.% NaCl solution to evaluate the surface electrochemical behavior and passivation characteristics of Al-1Y-xMg alloys with varying Mg contents (Fig. 9). Throughout the tests, all alloys exhibited some potential noise, yet the OCP values generally stabilized within ± 0.02 V, indicating a dynamic equilibrium between passive film formation and dissolution. As Mg content increased from 0 to 8 wt.%, the steady-state OCP progressively shifted in the negative direction, from approximately -0.72 V for the 0Mg alloy to -0.73 V, -0.74 V, -0.77 V, and -0.85 V for the 2Mg, 4Mg, 6Mg, and 8Mg alloys, respectively, reflecting the enhanced surface electrochemical activity induced by Mg addition. Notably, the 6Mg and 8Mg alloys displayed more pronounced high-frequency fluctuations in their OCP-time curves, which are commonly attributed to the repeated initiation and repassivation of metastable pits on the alloy surface.

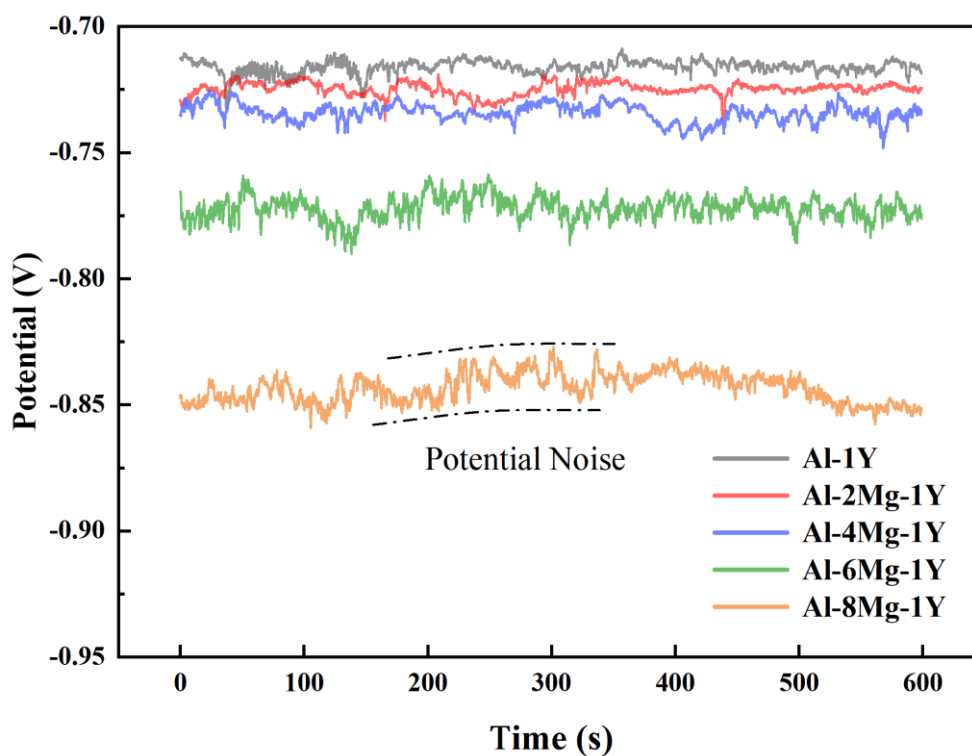


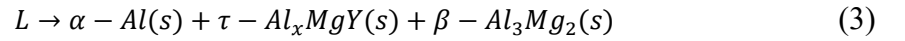
Fig. 9. Open circuit potential curves of as-cast Al-1Y-xMg alloys in 3.5 wt.% NaCl solution, showing the evolution of surface electrochemical stability with increasing Mg content.

4. Discussion

4.1 Mg-Controlled Solidification Pathways and Eutectic Network Formation

In order to understanding the effect of Mg variations on the microstructure of Al-1Y-xMg alloys, the nucleation sequence of the τ -Al_xMgY and β -Al₃Mg₂ phases must be considered as a primary factor. The synergistic effects of thermodynamics and kinetics are crucial, particularly under near-equilibrium cooling conditions. The formation of these two phases is governed not only by the alloy composition, especially the Mg content, but also by temperature and cooling rate, which ultimately determine the alloy's microstructure and macroscopic properties.

The phase diagram projection of the Al-Mg-Y system, particularly its liquidus projection, reveals its evolution with varying Mg content. Studies indicate that within the 1 wt.% Y section, as the Mg content increases from 0 wt.% to 8 wt.%, three primary eutectic valleys E₁, E₂, and E₃ can be observed [35,36]:



The eutectic reaction E₁ occurs at a temperature (T_{E1}) of approximately 640 °C. This reaction serves as the critical pathway for the formation of the Al₃Y phase at lower Mg contents. The eutectic reaction E₂ takes place at a temperature (T_{E2}) of about 605 °C. When the Mg content reaches a certain level, the τ -phase begins to form, replacing the Al₃Y phase. The eutectic reaction E₃ occurs at a temperature (T_{E3}) of approximately 520 °C. At higher Mg contents and lower temperatures, the β -phase also participates in the eutectic reaction. The DSC results show a main peak in the range of 670 - 632 °C and a shoulder peak at 520 ± 2 °C. The deviation between these measured temperatures and the aforementioned eutectic temperatures is less than 5 °C, indicating that the cooling rate used in this study (10 K s⁻¹) is close to equilibrium conditions. This close agreement allows for direct reference to the equilibrium reactions when analyzing the phase transformation processes.

Based on the aforementioned solidification pathway, the precipitation of τ -Al_xMgY and β -Al₃Mg₂ in Al-1Y-xMg alloys the precipitation sequence of τ and β is primarily governed by thermodynamic phase stability, with their sequential appearance following the descending temperature order of the eutectic reactions E₂ → E₃. However, the actual morphology, volume fraction, and spatial distribution of these phases are also influenced by kinetic factors, particularly the solute partitioning and diffusion kinetics during non-equilibrium solidification. Due to the lowest Gibbs free energy for τ -phase formation at this temperature, it preferentially nucleates at grain boundaries and rapidly develops a three-dimensional

skeleton. With progressive solidification, Y in the residual liquid is continuously depleted, while the Mg concentration relatively increases, causing the system to descend along the eutectic valley toward lower temperatures. When the composition reaches the ternary eutectic point, the remaining liquid undergoes an invariant reaction, leading to the cooperative precipitation of $\beta\text{-Al}_3\text{Mg}_2$ with the pre-existing τ -phase, forming a continuous network that fills the inter-dendritic regions.

The formation sequence of τ and β follows a temperature-decreasing order, governed by thermodynamic stability, resulting in their sequential precipitation. The prior nucleation of $\tau\text{-Al}_x\text{MgY}$ determines the solute redistribution path, thereby predefining the nucleation sites and timing for $\beta\text{-Al}_3\text{Mg}_2$. Fig. 10(a) presents a schematic diagram of the microstructure formation at different Mg contents, while Fig. 10(b-d) show the corresponding SEM images of the respective Mg content regions. The resulting microstructure, characterized by τ islands encased within a β network, is a direct outcome of this sequential solidification path. As shown in Fig. 10(c), the τ phase, which precipitates first along the grain boundaries, is surrounded by the β phase. However, during pretreatment, the β phase was etched away by Keller's reagent, leaving distinct etching pits around the τ phase.

Furthermore, the OCP results provide additional electrochemical evidence supporting the microstructural observations, particularly the dominant influence of the continuous $\beta\text{-Al}_3\text{Mg}_2$ network on the alloy surface state. At relatively low Mg contents, a dense and stable oxide film can form on the alloy surface, resulting in comparatively higher and more stable OCP values [37]. With increasing Mg content, $\beta\text{-Al}_3\text{Mg}_2$ precipitates continuously along the grain boundaries and gradually forms interconnected networks. Compared with the $\alpha\text{-Al}$ matrix, the β phase acts as an anodic constituent with higher electrochemical activity [38]. These continuous electrochemically active pathways compromise the integrity of the passive film and promote localized dissolution, leading to a significant negative shift in OCP. This interpretation is consistent with the intensified potential fluctuations observed in the 6Mg and 8Mg alloys, further indicating enhanced metastable pitting activity associated with the continuous $\beta\text{-Al}_3\text{Mg}_2$ network.

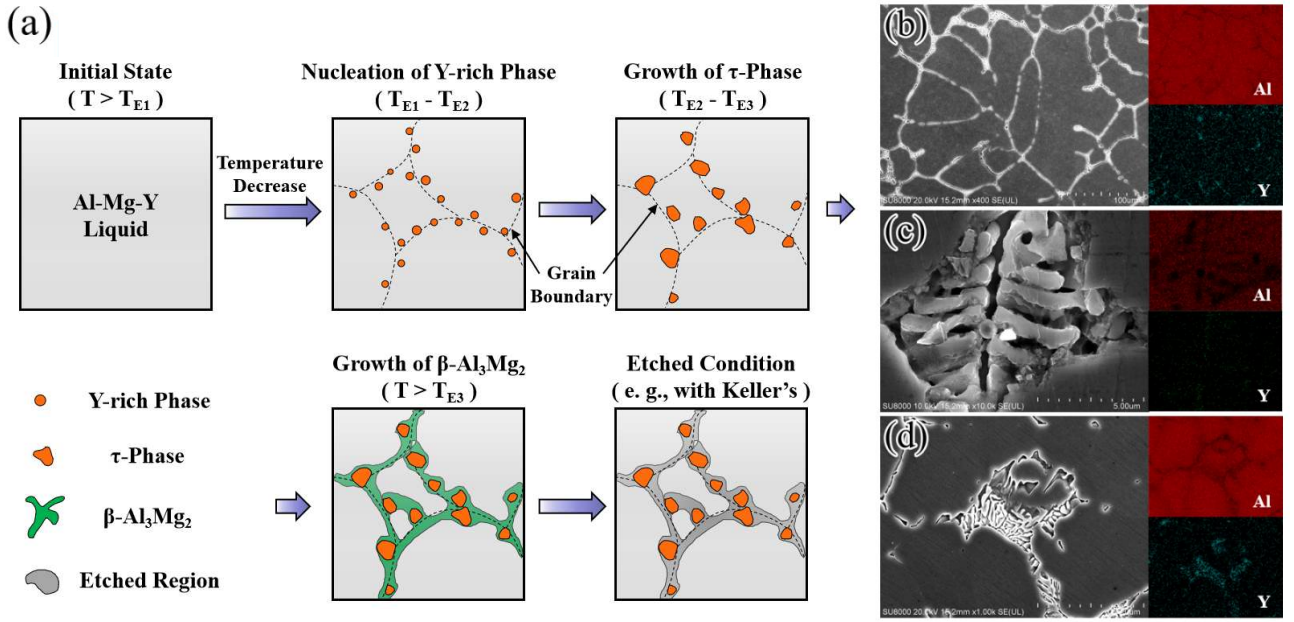


Fig. 10. (a) Schematic illustration of the sequential precipitation of τ -phase and β - Al_3Mg_2 phase and the resulting etched morphology, (b) SEM image corresponding to the Al-Y-2Mg alloy (c) SEM image corresponding to the Al-Y-4 Mg alloy, (d) SEM image corresponding to the Al-Y-6 Mg alloy.

4.2 Strengthening and Ductilization Mechanisms

The overall properties of the Al-1Y-xMg cast alloy are presented in the radar plot in Fig. 11(a). When the Mg content reaches ≥ 4 wt.%, both the yield strength and tensile strength of the alloy are significantly higher than those of samples without Mg or with low Mg content. Based on the solidification path discussed earlier, this strengthening increment is primarily attributed to the τ - $\text{Al}_x\text{MgY} + \beta$ - Al_3Mg_2 network formed through the E2 and E3 pseudo-ternary eutectic reactions, which effectively impedes dislocation motion. The strengthening mechanisms considered in this study mainly include grain refinement strengthening and precipitation strengthening. Assuming that their contributions to the overall strength are independent and generally considered additive, the total strength of the Al alloy in this work can be estimated using the following Eq. (3) [39,40]:

$$\sigma_{0.2} = \sigma_{GB} + \sigma_{prec} \quad \text{Eq. (3)}$$

In the equation, $\sigma_{0.2}$ represents the total yield strength; σ_{GB} denotes the contribution from grain boundary strengthening; σ_{prec} refers to the precipitation strengthening provided by τ phase and β phase. Grain refinement serves as an important strengthening mechanism by enhancing the resistance of grain boundaries to dislocation motion. This relationship can be described by the Hall-Petch Eq. (4) [41]:

$$\sigma_0^{-\frac{1}{2}}_{GB} \quad \text{Eq. (4)}$$

where σ_0 represents the yield stress of pure Al (20 MPa), k is a positive yield constant, and d denotes the average grain size. K is generally associated with the stress intensity of dislocation pile-up at grain

boundaries and typically ranges from 0.06 to 0.09 MPa·m^{1/2} [42]. When the Mg content reaches ≥4 wt.%, τ -Al_xMgY and β -Al₃Mg₂ phases precipitate along the grain boundaries.

According to the Orowan strengthening mechanism, the value of σ_{prec} can be estimated using the specific expression given in Eq. (5) [43]:

$$\sigma_{prec} = \frac{0.85Gb\ln(\frac{x}{b})}{2\pi(1-x)} \quad \text{Eq. (5)}$$

where G represents the shear modulus (G = 26 GPa), b denotes the Burgers vector (b = 0.286 nm), x is the average diameter, and l is the mean spacing. The discrepancy between the calculated results and experimental data may stem from the continuous or agglomerated β phase formed at grain boundaries when the Mg content becomes excessive. As the β phase lacks combination with the τ phase, its isolated presence at grain boundaries as a hard and brittle phase can lead to deterioration in alloy performance. Fig. 11(b) schematically illustrates the influence of precipitates with different compositions and structures on dislocation behavior.

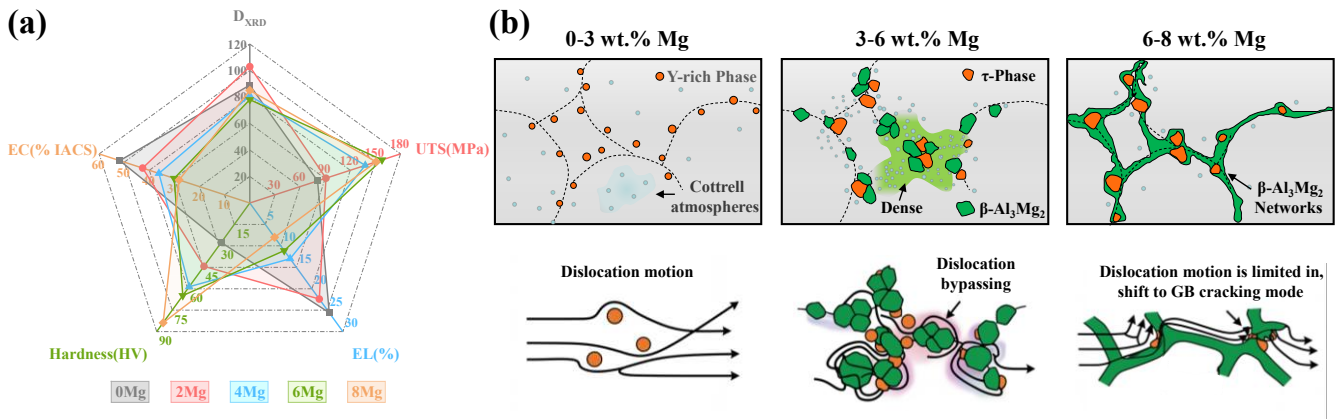


Fig. 11. (a) Radar plot showing the overall properties of as-cast Al-1Y-xMg alloys, (b) Influence of precipitates with different compositions and structures on dislocation behavior during deformation.

In Al-Mg-Y alloys, the type and distribution of precipitated phases are strongly dependent on Mg content, which in turn significantly affects dislocation motion and thus governs the mechanical properties. At low Mg concentrations (0-3 wt.%), the grain boundaries are mainly decorated with fine, dispersed Y-rich particles. During tensile deformation, dislocations bypass these particles via the Orowan mechanism. Due to the relatively large interparticle spacing, the strengthening contribution is limited, and the matrix remains largely continuous, resulting in high ductility but relatively low strength. At intermediate Mg levels (3-6 wt.%), the grain boundaries are dominated by agglomerates of β and Y-rich phases. Dislocations require higher applied stress to bypass these clusters, leaving dislocation loops around the precipitates. This increases alloy strength but reduces ductility. The enhanced obstruction to dislocation motion promotes local stress concentration, leading to a decrease in uniform elongation. At high Mg contents (6-8 wt.%), β

phases not only exist as agglomerates but also tend to form continuous networks. The reduced effective spacing within these networks increases the Orowan stress during dislocation bypass, causing severe local stress concentrations that readily initiate microcracks. Simultaneously, a substantial fraction of Mg atoms is consumed in β -Al₃Mg₂ precipitation, lowering the solute Mg concentration in the matrix and weakening solid-solution strengthening. Consequently, the tensile strength peaks at 6 wt.% Mg (155 MPa) and decreases to 147 MPa at 8 wt.% Mg, with a concurrent reduction in uniform elongation.

Room-temperature tensile tests further reveal typical Portevin-Le Chatelier (PLC) serrated flow behavior for alloys with Mg $\geq$ 2 wt.% (Fig. 12(a)), indicative of pronounced dynamic strain aging (DSA). The repeated pinning-depinning of dislocations associated with DSA obscures the macroscopic yield point [44]. To quantify this effect, the stress amplitude ($\Delta\sigma$) and the number of serrations (PLC bands) were analyzed, where $\Delta\sigma$ reflects the pinning strength of solute atoms or clusters and t_w denotes the pinning time at obstacles [45]. The frequency distribution of $\Delta\sigma$ for alloys with different Mg contents is summarized in Fig. 12(b), showing that both the serration amplitude and number initially increase with Mg content and then decline. At low Mg levels, grain boundaries are mainly dominated by Y-rich phases, while sparse Mg solutes form weak Cottrell atmospheres, requiring minimal stress for dislocation depinning. As a result, serrations are of low amplitude and relatively few in number, DSA is weak, and plastic deformation is dominated by uniform dislocation slip, yielding high elongation but limited solid-solution strengthening. With increasing Mg concentration, solute segregation around dislocations intensifies, enhancing pinning and increasing $\Delta\sigma$, which introduces additional dislocations, elevates work hardening, and raises the ultimate tensile strength. However, repeated localized deformation also promotes concentrated propagation of PLC bands, leading to ductility loss. When Mg content reaches 8 wt.%, the total number of serrations decreases to a level similar to low-Mg alloys, and the amplitude distribution shifts toward lower values. This behavior arises from microstructural evolution induced by excess Mg: continuous or semi-continuous networks of brittle β -Al₃Mg₂ form along grain boundaries, consuming a large fraction of Mg solutes and reducing their availability for DSA. Consequently, fewer mobile dislocations are present, PLC band nucleation is suppressed, and the deformation mechanism transitions from DSA-dominated plastic flow to brittle grain-boundary cracking. Stress concentration occurs preferentially at continuous β -phase networks, leading to early microcrack initiation rather than PLC serrations, which explains the observed decrease in ultimate tensile strength and the minimum elongation at 8 wt.% Mg.

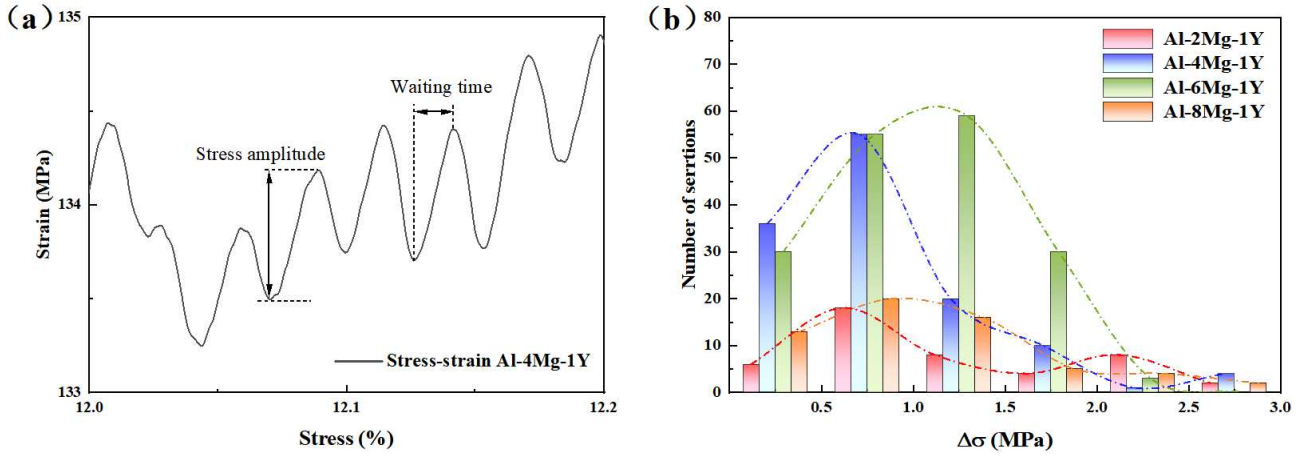


Fig. 12. (a) Schematic diagram of stress amplitude ($\Delta\sigma$) and waiting time (t_w); (b) Frequency distribution histograms of stress amplitude for samples with different Mg contents.

4.3 Electron and Phonon Scattering Mechanisms

With increasing Mg content, both electrical and thermal conductivities gradually decrease, which can be understood in terms of electron and phonon scattering mechanisms. In the Al-1Y-xMg alloy system, the total resistivity can be fully described according to the Mattissen-Flemming rule as shown in Eq. (6) [46]:

$$\rho = \rho^{\text{pure Al}} + S_{\text{GB}}\Delta\rho^{\text{GB}} + L^{\text{dislo}}\Delta\rho^{\text{vac}} + \sum_i c_{\text{soli}}^i \Delta\rho_{\text{soli}}^i \quad \text{Eq. (6)}$$

where $\rho^{\text{pure Al}}$ represents the resistivity of pure aluminum at room temperature, approximately 62% IACS [47] ($2.665 \times 10^{-8} \Omega \cdot \text{m}$). $\Delta\rho^{\text{GB}}$, $\Delta\rho^{\text{dislo}}$, $\Delta\rho^{\text{vac}}$ and $\Delta\rho_{\text{soli}}^i$ denote the contributions of grain boundaries, dislocations, vacancies, and solute element i to the resistivity, respectively. S_{GB} is the grain boundary area per unit volume, L_{dislo} is the dislocation density, C_v is the vacancy concentration, and C_i is the concentration of solute element i in the matrix.

However, in the present study, it is necessary to simplify Eq. (6) based on the specific microstructural characteristics of the alloys. First, all alloys were prepared in the as-cast condition without severe plastic deformation, and XRD line broadening analysis (Table 2) confirmed that the dislocation density remained on the order of 10^{14} m^{-2} with limited variation. Miyajima et al. [48] reported that the resistivity contribution of dislocations, $\Delta\rho^{\text{dislo}}$, is on the order of $2.7 \times 10^{-25} \Omega \cdot \text{m}^3$. Accordingly, the resistivity increment introduced by dislocations in this study does not exceed $0.003 \times 10^{-8} \Omega \cdot \text{m}$, which is negligible compared with the contributions from solute atoms and precipitates. Secondly, the vacancy concentration at room temperature under equilibrium conditions is extremely low, and its contribution to resistivity can also be neglected. Moreover, optical microstructural observations show that variations in Mg content do not significantly affect the α -Al grain size, indicating that the $\Delta\rho^{\text{GB}}$ term remains essentially constant across the alloys and can be regarded as part of the baseline resistivity. Therefore, Eq. (6) can be reasonably simplified as:

$$\rho_{\text{total}} = \rho^{\text{pure Al}} + \Delta\rho^{\text{GB}} + \Delta\rho^{\text{soli}} + \Delta\rho^{\text{prec}} \quad \text{Eq. (7)}$$

where $\rho^{\text{pure Al}}$ and $\Delta\rho_{\text{GB}}$, representing the baseline contribution from pure aluminum and grain boundary scattering; $\Delta\rho^{\text{solu}}$ primarily reflects the scattering effect of solute Mg atoms; and $\Delta\rho^{\text{prec}}$ represents the electron scattering contribution at interfaces of precipitates such as $\tau\text{-Al}_x\text{MgY}$ and $\beta\text{-Al}_3\text{Mg}_2$. Eq. (7) focuses the mechanisms of resistivity variation in this alloy system on two key variables-the solute Mg atoms and the secondary precipitates. The following discussion examines their relative contributions in different Mg content ranges.

The equilibrium solubility of Mg in $\alpha\text{-Al}$ at room temperature is limited. Although supersaturated solid solutions can be obtained in the present as-cast alloys due to non-equilibrium solidification, the amount of Mg that can be retained in the matrix still has a practical upper limit. Therefore, $\Delta\rho^{\text{solu}}$ does not increase linearly with nominal Mg content but tends to saturate once the solubility limit is reached. Excess Mg beyond this limit mainly precipitates as $\beta\text{-Al}_3\text{Mg}_2$ and related phases; at this stage, the further increase in $\Delta\rho^{\text{total}}$ should be primarily attributed to the increase in $\Delta\rho^{\text{prec}}$.

The $\Delta\rho^{\text{total}}$ of the 2Mg alloy is $0.71 \times 10^{-8} \Omega \cdot \text{m}$. According to Gaber et al. [49], who investigated solid-solution Al-Mg alloys, the resistivity increment caused by electron-impurity scattering is proportional to the atomic fraction of solute Mg, with approximately 12.99% contribution per 1 at % Mg. Based on this relationship, the theoretical $\Delta\rho^{\text{solu}}$ of the 2Mg alloy under a fully solid-solution state is approximately $0.78 \times 10^{-8} \Omega \cdot \text{m}$, which agrees well with the measured $\Delta\rho^{\text{total}}$. This indicates that Mg is almost entirely in solid solution at this stage, with $\Delta\rho^{\text{total}} \approx \Delta\rho^{\text{solu}}$ and a negligible contribution from $\Delta\rho^{\text{prec}}$.

With further increase in Mg content, the solubility limit of Mg is gradually reached, and microstructural observations reveal the precipitation of $\tau\text{-Al}_x\text{MgY}$ and $\beta\text{-Al}_3\text{Mg}_2$ phases along grain boundaries. It can thus be inferred that the concentration of solute Mg in the matrix approaches saturation at 4Mg, and the upper limit of $\Delta\rho^{\text{solu}}$ is approximately $1.44 \times 10^{-8} \Omega \cdot \text{m}$, beyond which it no longer increases significantly with Mg content.

In the high Mg content range (6-8 wt.%), the $\beta\text{-Al}_3\text{Mg}_2$ phase exists not only as discrete clusters but also tends to form semi-continuous or fully continuous networks along grain boundaries. This microstructural evolution is schematically illustrated in Fig. 13, and Table 6 summarizes the estimated values of $\Delta\rho^{\text{solu}}$ and $\Delta\rho^{\text{prec}}$ derived from the above analysis.

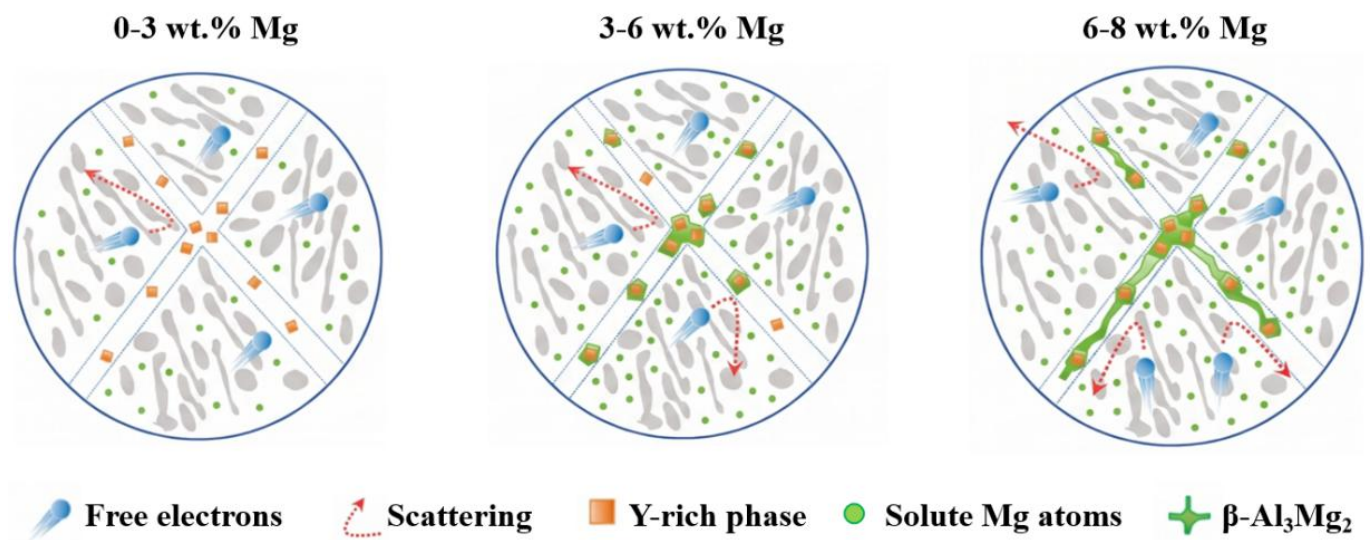


Fig. 13. Schematic diagram of free electron transport in regions with different Mg contents.

Table 6 Measured conductivity, resistivity, and scattering mechanism analysis for Al-xMg-Yalloys

Alloy	Measured conductivity (%IACS)	ρ_{total} ($\times 10^{-8} \Omega \cdot m$)	$\Delta \rho^{total}$ ($\times 10^{-8} \Omega \cdot m$)	$\Delta \rho^{solu}$ ($\times 10^{-8} \Omega \cdot m$)	$\Delta \rho^{prec}$ ($\times 10^{-8} \Omega \cdot m$)
Al-1Y	51.80 ± 0.70	3.33	—	—	—
Al-2Mg-1Y	42.77 ± 0.53	4.04	0.71	≈ 0.71	≈ 0
Al-4Mg-1Y	36.17 ± 0.37	4.77	1.44	≈ 1.13	≈ 0.31
Al-6Mg-1Y	30.12 ± 0.30	5.75	2.42	≈ 1.36	≈ 1.06
Al-8Mg-1Y	28.16 ± 0.03	6.12	2.79	≈ 1.42	≈ 1.37

Notably, the OCP measurements provide independent electrochemical evidence supporting the resistivity analysis, as both phenomena share the same microstructural origin. OCP reflects the integrity of the surface passive film and surface electrochemical activity, while resistivity gauges the transport capacity of free electrons within the bulk material. When $Mg \geq 6$ wt.%, the continuous β - Al_3Mg_2 network along the grain boundaries negatively affects alloy performance in two ways: electrochemically, the β phase acts as an anodic constituent relative to the α -Al matrix, forming preferential electrochemical pathways that compromise the passive film, causing OCP to shift significantly from approximately -0.72 V in 0Mg to -0.85 V in 8Mg, accompanied by intensified high-frequency potential fluctuations and increased metastable pitting events; structurally, this network creates a high density of phase interfaces and extended scattering paths, greatly impeding electron transport, thereby increasing resistivity and simultaneously decreasing thermal conductivity.

5 Conclusions

This study quantitatively links Mg-controlled solidification pathways to the microstructure and properties of as-cast Al-1Y-xMg alloys. The principal conclusions are summarized as follows:

(1) As the Mg content increases from 0 to 8 wt.%, the solidification pathway of as-cast Al-1Y-xMg alloys sequentially evolve through three distinct stages: isolated Y-rich phase $\rightarrow$ τ -Al_xMgY precipitation $\rightarrow$ τ + β eutectic network. In the 6-8 wt.% Mg range, a continuous τ + β grain-boundary eutectic network forms, representing a highly interconnected microstructural morphology achievable through compositional control during casting. This network topology serves as the structural basis for subsequent mechanical and physical property modulation.

(2) When the Mg content reaches 6 wt.%, the tensile strength of the alloy reaches a peak value of 158.0 MPa, representing an increase of approximately 95% compared with the Mg-free alloy. This improvement is primarily attributed to solid-solution strengthening from Mg and precipitation strengthening from τ / β phases. Meanwhile, the uniform elongation remains at 11.0% at this composition, and the fracture mode is dominated by ductile dimples, indicating a favorable strength-ductility synergy achievable through controlled solidification.

(3) Quantitative analysis of Portevin-Le Chatelier (PLC) serrations reveals that Mg content governs the competition between solute Cottrell atmospheres and brittle β -network embrittlement. At intermediate Mg levels (4-6 wt.%), dynamic strain aging enhances work hardening, contributing to strength improvement. At 8 wt.% Mg, excessive β -phase network formation consumes solute Mg, suppresses PLC activity, and shifts the deformation mechanism from dislocation-solute interaction to grain-boundary cracking, accounting for the observed ductility loss.

(4) By varying the Mg content—and thus the eutectic network topology—the electrical conductivity of the alloy can be continuously tuned within 28.2-51.8% IACS, while the thermal conductivity can be adjusted in the range of 148-218 W·m⁻¹·K⁻¹. Mechanistic analysis based on the Matthiessen-Flemming rule indicates that at low Mg contents, electron scattering is primarily governed by solute Mg atoms, whereas at high Mg contents (≥ 6 wt.%), τ / β phase-interface scattering gradually becomes dominant. This transition directly correlates with the evolution of the eutectic network during solidification.

(5) This study establishes that Mg content serves as a powerful control variable during casting to tailor the solidification pathway, eutectic network morphology, and resultant mechanical and physical properties of Al-1Y-xMg alloys. The quantitative relationships identified herein provide a design guideline for cast Al-Mg-Y alloys targeting combined structural and functional applications.

Funding

The work was supported by the Guangxi Natural Science Foundation (2025GXNSFAA069657), the Special Fund for Science and Technology Development of Guangxi (AD25069078), and the National Natural Science Foundation of China (52261024).

CRedit authorship contribution statement

Ren Yulong: Writing-original draft, Investigation, Data curation. **Yang Xuexuan:** Investigation. **Tang Peng:** Writing-review & editing, Resources, Project administration, Investigation, Funding acquisition.

Data Availability

Data will be made available on request.

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