

Article

Microstructural Constituents, Local Hardness Contrast, and Tensile Variability of As-Cast AM50 Magnesium Alloy

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Abstract

Commercially produced as-cast AM50 magnesium alloy was investigated to assess the relationship between cast microstructural heterogeneity, local hardness contrast, tensile variability, and fracture behavior. Optical microscopy (OM), scanning electron microscopy coupled with energy-dispersive X-ray spectroscopy (SEM/EDS), X-ray diffraction (XRD), semiquantitative image analysis, hardness testing, tensile testing, and fractography were employed. The microstructure was dominated by an α -Mg matrix containing interdendritic Al-enriched constituents consistent with β -Mg₁₇Al₁₂ and Al-Mn-rich particles. XRD confirmed α -Mg and β -Mg₁₇Al₁₂, whereas the Al-Mn-rich constituents were supported by SEM/EDS but could not be reliably indexed by XRD. Image analysis of eight optical fields yielded an apparent area fraction of $27.6 \pm 1.1\%$ for dark interdendritic/secondary regions; this value represents combined etched microstructural constituents rather than an individual phase fraction. Vickers microhardness showed a pronounced local contrast between the matrix (39.1 ± 5.3 HV) and constituent-rich regions (119 ± 12 HV). Tensile testing of six specimens extracted from comparable upper regions of different commercial ingots yielded an average ultimate tensile strength of 129.42 ± 17.63 MPa, while recorded fracture strain showed substantially greater relative scatter (coefficient of variation $\approx 122\%$) than strength. Fractography revealed mixed ductile–brittle features, including dimples, tear ridges, and localized brittle or rough regions. The combined results show that intermetallic-rich regions contribute to local mechanical heterogeneity, but their presence alone is insufficient to explain the tensile variability. Casting-related discontinuities, sampling history, and between-ingot variability must also be considered.

Keywords: AM50 magnesium alloy; cast microstructure; intermetallic constituents; local hardness contrast; tensile variability; fractography



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1. Introduction

Magnesium alloys have gained increasing attention as lightweight structural materials due to their low density, high specific strength, and excellent energy absorption capacity, making them attractive for transportation and protection applications [1]. In the automotive, railway, aerospace, and defense sectors, these alloys enable weight reduction and

lower fuel consumption while providing adequate mechanical performance and favorable impact behavior [2,3]. In ballistic and armor-related applications, magnesium combines low density, high damping capacity, and good shock absorption, making it a potential candidate for lightweight armor systems when its microstructure and texture are properly optimized [4]. Furthermore, several studies have demonstrated that magnesium alloy structures and tubes can achieve high specific energy absorption under impact and crushing conditions, reinforcing their potential for impact-resistant components in vehicles, trains, and aerospace systems [5–7].

The hexagonal close-packed (HCP) crystal structure of magnesium limits the number of easily activated slip systems at room temperature, resulting in low ductility and formability in many magnesium alloys [8]. The competition and interaction between slip and twinning mechanisms, controlled by factors such as texture, temperature, alloying elements, and critical resolved shear stresses, determine the plastic behavior, anisotropy, and the balance between strength and ductility of these alloys [9].

AM50 magnesium alloys, belonging to the Mg–Al–Mn system, have been extensively studied as low-density materials with microstructures that can be tailored through alloying additions and heat treatments [10]. The typical as-cast microstructure consists of an α -Mg matrix, β -Mg₁₇Al₁₂ phase, and Al_xMn_y particles, which may be refined or modified through the addition of elements such as Sb, Cu, Ce, Si, or rare-earth elements, significantly affecting strength, ductility, hardness, and wear behavior [11–13]. These microstructural modifications translate into attractive mechanical properties, including good ductility, impact energy absorption capability, and strength levels suitable for lightweight structural components, although AM50 generally exhibits lower overall performance than some other magnesium alloys in material selection rankings [7]. Due to its low density, good castability, and responsiveness to microstructural modifications, as-cast AM50 is widely employed and considered for automotive safety components and impact-absorbing structures manufactured by high-pressure die casting and related processes [13,14].

The fracture behavior of AM50 magnesium alloy is strongly influenced by its cast microstructure, including intermetallic particles, internal porosity, and secondary phases located along grain boundaries [12]. Crack initiation typically occurs at local discontinuities such as coarse voids, intermetallic particles, or inclusions, which promote stress concentration and microvoid formation under static or fatigue loading conditions [15]. Fractographic studies of AM50 and related alloys have revealed fracture paths ranging from quasi-cleavage brittle fracture to mixed ductile-brittle modes, reflecting the influence of porosity, intermetallic phase morphology, and thermal or surface treatments [16]. Therefore, the relationship between microstructure and mechanical properties in AM50 is governed by grain refinement, the nature and distribution of intermetallic phases, and internal defects, which collectively control strength, fatigue behavior, and fracture resistance [12,17].

This study provides an integrated characterization of an as-cast AM50 alloy through OM, SEM/EDS, XRD, semiquantitative image analysis, hardness testing, tensile testing and fractography. The objective is to characterize the local mechanical contrast between the α -Mg matrix and Al/Mn-enriched regions and to assess whether the observed distribution of these constituents is sufficient to explain the tensile variability. The interpretation explicitly considers the established sensitivity of AM50 to casting route, section thickness, sampling position and localized porosity.

2. Materials and Methods

2.1. Material and Metallographic Preparation

The material investigated in this study consisted of commercially produced as-cast AM50 magnesium alloy ingots supplied by an industrial manufacturer. According to the

manufacturer's specification, the alloy composition consisted of 4.91 wt.% Al, 0.04 wt.% Zn, 0.30 wt.% Mn, 0.0029 wt.% Cu, 0.03 wt.% Si, 0.0012 wt.% Fe, 0.0010 wt.% Ni, 0.0005 wt.% Be, and balance Mg. The ingots were produced using the manufacturer's standard industrial casting route. Detailed melt-handling and protection parameters were proprietary and were not available for independent assessment; therefore, the possible contribution of processing-related inclusions, oxide films, or other melt-handling-related discontinuities could not be directly evaluated.

The commercial ingots had a trapezoidal cross-section, with the upper surface exposed during solidification and the lateral and lower surfaces in contact with the mold. Consequently, heat-extraction conditions were expected to vary spatially within each ingot. All tensile specimens were machined from comparable positions in the upper region of different commercial ingots produced by the same nominal industrial route. Metallographic specimens used for OM, SEM/EDS, and XRD were obtained from representative material from the same upper region. This sampling strategy minimized within-ingot positional differences, although between-ingot variability in local solidification history and defect population cannot be excluded.

The metallographic specimen was hot mounted in phenolic resin and subsequently ground using waterproof SiC abrasive papers with progressively finer grit sizes. Polishing was then performed using diamond suspensions until a mirror-like surface suitable for metallographic examination was obtained.

Microstructural etching was performed using a glycol-based reagent composed of glycol, nitric acid (HNO_3), and acetic acid (CH_3COOH). The etching time was approximately 30 s, allowing adequate contrast between the microstructural constituents. The metallographic preparation and chemical etching procedures followed the laboratory protocol based on conventional metallographic practices for magnesium alloys (ASTM E407-23) [18].

2.2. Optical Microscopy (OM), Scanning Electron Microscopy (SEM) and Energy-Dispersive X-Ray Spectroscopy (EDS)

The microstructural characterization was initially performed by OM. Images were acquired before and after chemical etching to evaluate the surface condition and reveal the microstructural constituents present in the alloy using an Olympus BX53M optical microscope (Olympus Corporation, Tokyo, Japan).

Complementary microstructural analyses were carried out using a FEI Quanta FEG-250 field-emission scanning electron microscope (SEM) (FEI Company, Hillsboro, OR, USA). The observations were performed at an accelerating voltage of 20 kV using a field-emission gun (FEG) electron source. High-resolution micrographs were obtained to evaluate the morphology and distribution of the microstructural constituents. SEM/EDS analyses were performed on ten representative fields acquired from the same upper-base region of the AM50 ingot used for optical microscopy and tensile specimen extraction.

Chemical characterization was conducted using an energy-dispersive X-ray spectroscopy (EDS) detector coupled to the SEM. Point analyses and elemental mapping were performed to identify the chemical composition of the matrix and intermetallic phases present in the alloy. The EDS maps were acquired with a 50 μm scale bar and were used to evaluate the spatial distribution of Mg, Al, and Mn and to support the chemical identification of the secondary constituents observed in the optical micrographs. Due to the smaller analyzed area of the EDS fields compared with optical microscopy, EDS mapping was used mainly for the chemical interpretation of the intermetallic-rich regions, whereas optical microscopy was used for the estimation of the apparent area fraction of dark interdendritic/secondary constituents.

2.3. X-Ray Diffraction (XRD)

Phase identification was performed by XRD using a PANalytical Empyrean diffractometer (PANalytical, Almelo, The Netherlands). The equipment was operated with CuK α radiation ($\lambda = 1.5406 \text{ \AA}$), at 40 kV and 40 mA.

The diffraction patterns were collected to identify the crystalline phases present in the AM50 alloy and determine the diffraction peaks associated with their characteristic crystallographic planes. The measurements were performed in line-focus configuration using a step size of 0.03° , a counting time of 150 s per step, and a PIXcel detector with 255 channels.

The crystalline phases were identified by comparison with reference crystallographic data from the Crystallography Open Database (COD) and the International Centre for Diffraction Data (ICDD) Powder Diffraction File (PDF) database using HighScore Plus software (version 3.0e/3.0.5). The α -Mg phase was indexed using COD card No. 96-901-3055, whereas the β -Mg₁₇Al₁₂ phase was identified using ICDD/PDF card No. 01-073-1148.

2.4. Semiquantitative Image Analysis of Interdendritic/Secondary Constituents

The area fraction of dark interdendritic/secondary constituents was estimated by digital image analysis of optical micrographs acquired after chemical etching. The analysis was performed using ImageJ software (version 1.54g) on eight non-overlapping optical microscopy fields obtained from the upper-base region of the AM50 ingot, corresponding to the same region from which the tensile specimens were extracted.

All optical micrographs were acquired at the same magnification and contained a 200 μm scale bar. Based on the calibrated image dimensions, each field corresponded to an analyzed area of approximately 2.2 mm^2 , resulting in a total analyzed area of approximately 17.6 mm^2 for the eight fields.

Before image processing, scale bars, black borders, and non-microstructural regions were excluded from the analysis. The images were converted to grayscale, and a consistent thresholding procedure was applied to separate the darker microstructural constituents from the lighter α -Mg matrix. The measured values are reported as the apparent area fraction of dark interdendritic/secondary constituents.

The chemical nature of the dark interdendritic/secondary regions was evaluated separately by SEM/EDS point analyses and elemental mapping, which indicated localized Al- and Mn-enrichment. XRD independently confirmed the presence of β -Mg₁₇Al₁₂, while the Al-Mn-rich constituents observed by SEM/EDS could not be reliably indexed by XRD because of their low fraction and possible peak overlap. Therefore, the optical area fraction represents the combined apparent fraction of dark interdendritic/secondary constituents and should not be interpreted as the fraction of a single intermetallic phase.

2.5. Hardness Testing

2.5.1. Vickers and Knoop Microhardness

Vickers (HV) and Knoop (HK) microhardness tests were performed according to ASTM E384-22 [19] using a Shimadzu HMV-G microhardness tester (Shimadzu Corporation, Kyoto, Japan). For both tests, a load of 50 gf was applied, and ten indentations were performed for each hardness scale to ensure statistical representativeness. The indentation dimensions were determined using the optical measurement system integrated into the equipment, providing micrometric precision. The reported hardness values correspond to the average of ten measurements.

2.5.2. Rockwell F and Brinell Hardness

Rockwell hardness measurements were performed according to ASTM E18-22 [20] using a Wilson Rockwell 574 Hardness Tester (Wilson Instruments, Norwood, MA, USA), operating on the Rockwell F (HRF) scale. A load of 60 kgf was applied, and ten indentations were performed. Brinell hardness (HB) tests were conducted according to ASTM E10-23 [21] using an Amsler Otto Wolpert-Werke Diatestor 751 hardness tester (Amsler Otto Wolpert-Werke, Ludwigshafen, Germany). A load of 187.5 kgf was applied, and ten indentations were performed. For both hardness methods, the reported values correspond to the average of ten measurements.

2.6. Tensile Test and Fractographic Analysis

Six tensile specimens were machined from comparable upper regions of different commercial AM50 ingots according to ASTM E8/E8M-24 [22] specifications for subsize rectangular specimens, as shown in Figure 1a. The machining procedure followed the dimensional tolerances established by the standard.

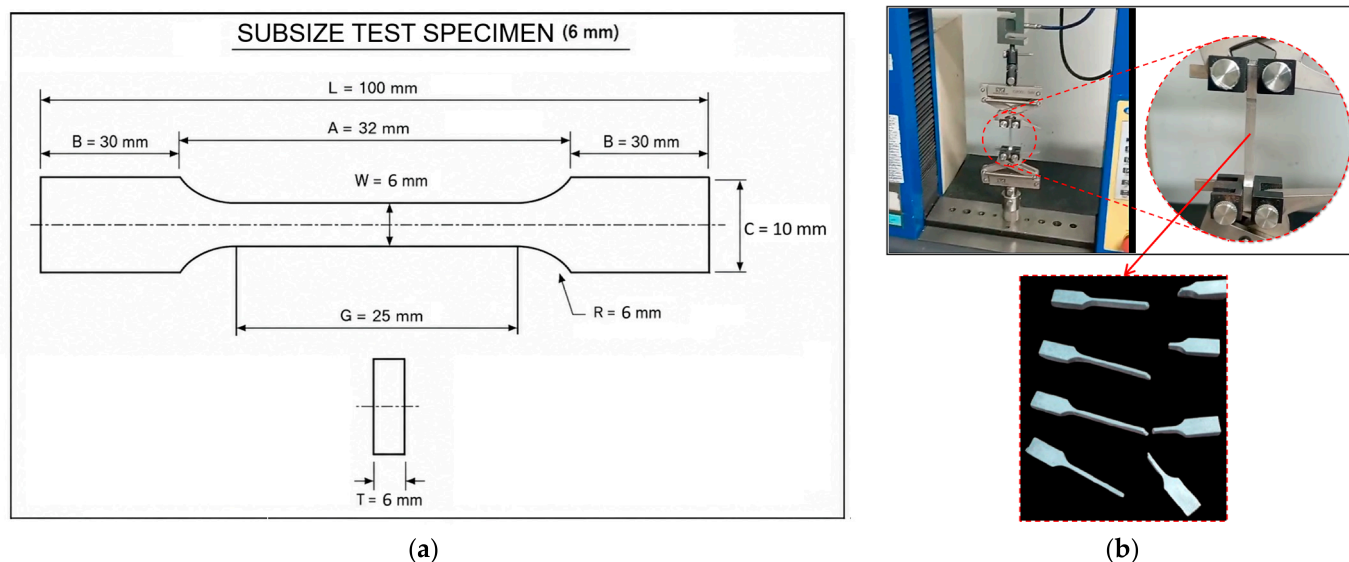


Figure 1. Geometric and experimental characterization of the material: (a) technical drawing detailing the nominal dimensions (in mm) of the subsize specimen used and (b) experimental apparatus for the tensile test, with an enlarged detail of the specimen properly fixed to the grips before load application, and specimens after the test. In (a), L represents the total specimen length, B the length of the grip sections, A the length of the reduced section, W the width of the reduced section, G the gauge length, R the fillet radius, C the width of the grip sections, and T the specimen thickness. The red dashed lines, circles, and arrows indicate the regions highlighted and enlarged for detailed visualization.

The tensile tests were carried out using an EMIC DL-1000 Universal Testing Machine (EMIC, São José dos Pinhais, Paraná, Brazil) equipped with a 10 kN load cell. Strain measurements were obtained using a contact extensometer attached to the gauge section of the specimens. Tests were performed at room temperature under uniaxial loading conditions with continuous acquisition of load and displacement data, allowing the construction of engineering stress-strain curves (Figure 1b).

After tensile testing, the fracture surfaces of the specimens were examined by SEM (FEI Quanta FEG-250) to identify the dominant fracture mechanisms. The analyses focused on the characterization of fracture features such as cleavage facets, dimples, secondary cracks, pores and fractured intermetallic particles, enabling the correlation between fracture behavior, microstructure and mechanical properties of the AM50 alloy.

3. Results and Discussion

3.1. Microstructural Characterization and Phase Identification

The microstructural analyses conducted by optical microscopy (OM) (Figure 2a–c) and scanning electron microscopy (SEM) revealed that the AM50 alloy exhibits the characteristic microstructure of cast Mg–Al–Mn alloys. The microstructure is predominantly composed of an α -Mg matrix accompanied by heterogeneously distributed secondary phases. After chemical etching, etched microstructural boundaries and interdendritic features were revealed, together with precipitates and segregated regions concentrated mainly at interdendritic interfaces and grain boundaries.

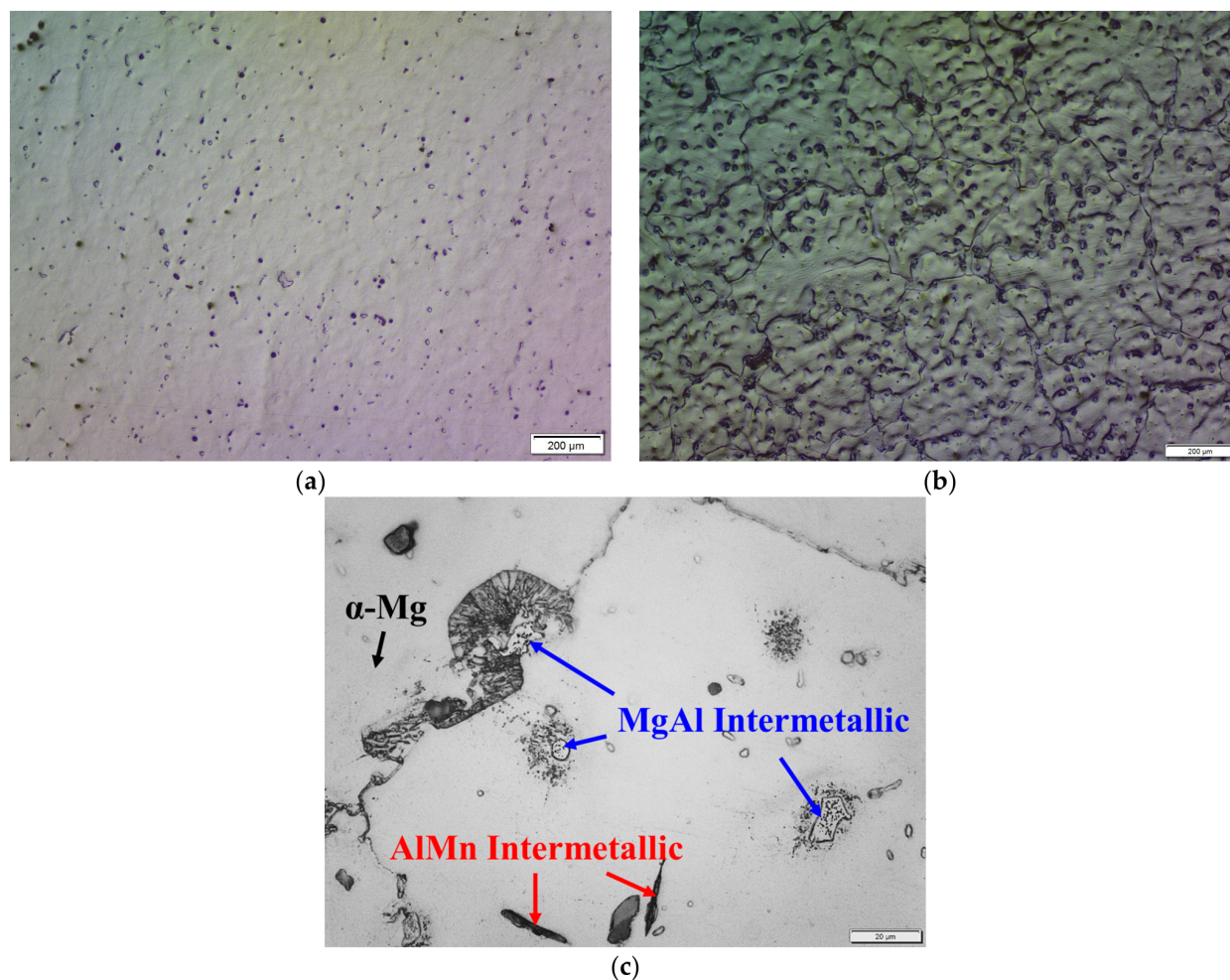


Figure 2. OM of AM50 magnesium alloy: (a) polished surface without chemical etching; (b) microstructure after chemical etching; and (c) higher-magnification detail highlighting the magnesium-rich matrix (α -Mg) and Mg–Al- and Al–Mn-rich secondary constituents.

Macroscopic etching attempts were performed on the sectioned ingot surface in order to evaluate the presence of solidification macrozones. However, the etched surface did not reveal clearly distinguishable macrostructural zones, such as columnar or equiaxed regions. Therefore, the macrostructural examination was used mainly to document the sampling region, while the interpretation of solidification features was based on optical microscopy and SEM/EDS analyses.

The optical micrographs obtained from different fields of the upper-base region of the AM50 ingot revealed a consistent heterogeneous cast microstructure. In the polished condition (Figure 2a), the microstructure exhibited a relatively homogeneous light-gray matrix, whereas after chemical etching (Figure 2b), the microstructural contrast became

more pronounced, allowing the identification of darker interconnected regions distributed throughout the matrix. These regions were preferentially located along interdendritic boundaries and are consistent with segregated areas and secondary constituents formed during solidification.

Although local variations in the distribution and morphology of the dark constituents were observed among the analyzed fields, the overall microstructural pattern was similar, indicating that the selected region presents a representative cast morphology. The presence of irregular and locally elongated dark regions in Figure 2b suggests solute redistribution during solidification and supports the subsequent SEM/EDS observations of Al- and Mn-enriched constituents.

The higher-magnification image shown in Figure 2c provides greater detail of the constituent phases, allowing the distinction between the Mg-rich α -Mg matrix and the secondary phases distributed predominantly in interdendritic regions. The observed morphology is consistent with the formation of Al- and Mn-containing constituents during solidification.

The optical images provided adequate contrast for qualitative characterization of the cast microstructure and for identifying regions of microstructural heterogeneity. However, they were not used for standardized grain-size measurement, since the revealed features are mainly associated with dendritic/interdendritic morphology and secondary constituent distribution rather than unequivocal crystallographic grain boundaries.

OM observations indicated that the cast AM50 alloy contains interdendritic regions enriched with secondary phases, exhibiting irregular microstructural/dendritic morphology and a heterogeneous distribution of microstructural constituents. This behavior is consistent with literature reports for AM50 alloys solidified without subsequent thermomechanical treatments [23,24], highlighting the influence of the solidification process on the resulting microstructure.

The SEM micrographs and EDS analyses indicated localized regions enriched in Al and Mn. The Al-enriched constituents are consistent with the β -Mg₁₇Al₁₂ phase (Figure 3) subsequently confirmed by XRD, whereas the Mn-enriched particles are consistent with Al-Mn-rich intermetallic constituents commonly reported in AM50, including Al₈Mn₅ (Figure 4). These constituents exhibited irregular morphologies and non-uniform distributions within the predominantly α -Mg matrix.

The SEM/EDS maps acquired from ten representative fields provided chemical support for the interpretation of the optical micrographs. The Mg maps showed a predominant Mg-rich matrix, while the Al and Mn maps revealed localized enriched regions associated with the darker interdendritic/secondary constituents observed by optical microscopy. These results indicate that the dark regions revealed by optical etching are chemically heterogeneous and are associated with Al- and Mn-enriched constituents formed during solidification.

Although the EDS maps were acquired over smaller fields than the optical micrographs, they provided higher chemical specificity and supported the preferential localization of Al- and Mn-enriched regions within the cast microstructure. Therefore, EDS mapping was used as a complementary technique to identify the chemical nature of the secondary constituents, while the optical micrographs provided a broader field for estimating their apparent area fraction.

These secondary constituents can influence the local mechanical response because their morphology, continuity, and spatial distribution create mechanical contrast with the α -Mg matrix. In Mg-Al alloys, β -Mg₁₇Al₁₂ and other harder secondary constituents may restrict local plastic accommodation and promote stress concentration at matrix/constituent

interfaces [25]. However, the present data do not establish that these constituents alone control the macroscopic tensile response.

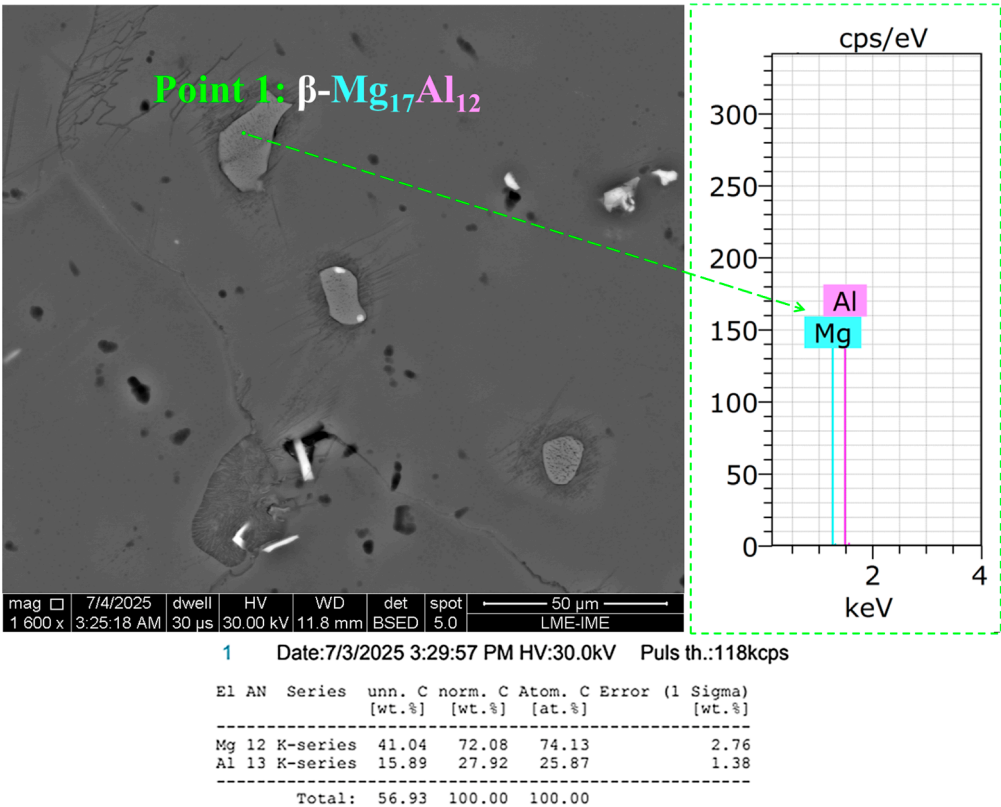


Figure 3. SEM image and point EDS analysis of a Mg-Al-rich constituent consistent with the β -Mg₁₇Al₁₂ phase identified by XRD. The green dashed rectangle indicates the enlarged EDS spectrum, while the green dashed line identifies the location of the analyzed point. The EDS spectrum shows the characteristic Mg and Al peaks associated with the β -Mg₁₇Al₁₂ phase.

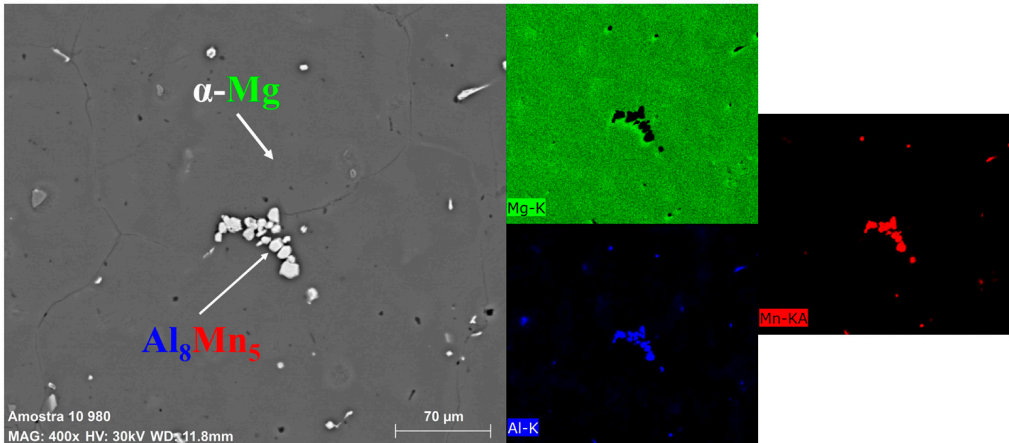


Figure 4. SEM image and individual EDS elemental maps showing an Al-Mn-rich constituent consistent with Al₈Mn₅. The green, red, and blue maps correspond to Mg, Mn, and Al, respectively, highlighting the elemental distribution within the analyzed region. The white arrow indicates the α -Mg matrix, while the colored label identifies the Al₈Mn₅-rich constituent.

The X-ray diffraction (XRD) pattern of the AM50 alloy (Figure 5) revealed the predominance of the α -Mg phase, which constitutes the alloy matrix. The corresponding diffraction peaks were indexed based on Crystallography Open Database (COD) card No. 96-901-3055.

Reflections corresponding to the (100), (002), (101), (102), (110), (103), (112), (201), (004), and (202) crystallographic planes were identified throughout the analyzed 2θ range.

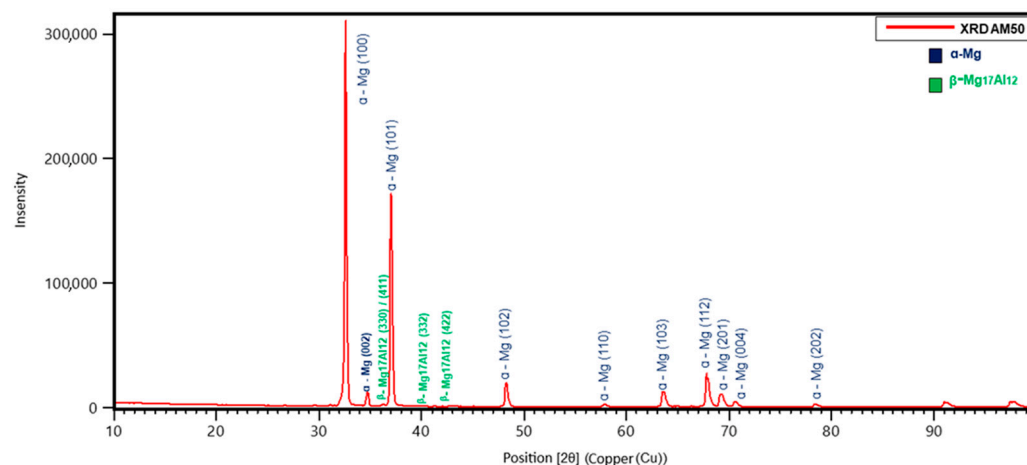


Figure 5. XRD pattern of the AM50 alloy.

In addition to the α -Mg matrix, lower-intensity diffraction peaks compatible with the β -Mg₁₇Al₁₂ intermetallic phase were observed and indexed using ICDD/PDF card No. 01-073-1148, confirming the presence of this secondary phase in the alloy microstructure.

Although Al–Mn-rich intermetallic phases are expected in the AM50 alloy, their identification in the XRD pattern was not possible. Their low fraction in the microstructure may result in diffraction reflections of insufficient intensity for reliable indexing, which may also overlap with reflections from the major α -Mg phase. Nevertheless, Al–Mn-rich particles were observed in the SEM/EDS analyses, supporting the presence of Al–Mn-rich constituents in the microstructure, although their low fraction prevented their identification by XRD.

The lower intensity of the β -Mg₁₇Al₁₂ reflections is consistent with the relatively low fraction of secondary constituents compared with the predominant α -Mg matrix. This observation is also consistent with the SEM/EDS characterization and the semiquantitative image analysis, which revealed a predominantly Mg-rich matrix with localized Al- and Mn-enriched regions.

The XRD results therefore provide evidence for the coexistence of the α -Mg matrix and β -Mg₁₇Al₁₂ intermetallic phase. Such microstructural heterogeneity is typical of cast AM50 alloys without thermomechanical treatment and significantly influences local properties such as hardness and tensile strength [23,24]. The Al–Mn-rich constituents identified by SEM/EDS could not be independently supported by XRD because of their low fraction and the resulting limitations in diffraction peak detection and indexing. These results complement the microstructural observations obtained by OM and SEM/EDS and contribute to the interpretation of the heterogeneous microstructure and its relationship with the mechanical behavior of the AM50 alloy.

3.2. Semiquantitative Analysis of Interdendritic/Secondary Constituents

Digital image analysis of the optical micrographs indicated that the dark interdendritic/secondary constituents occupied an average apparent area fraction of approximately $27.6 \pm 1.1\%$ of the analyzed region. The individual values ranged from approximately 26.3% to 29.8%, indicating a relatively consistent distribution among the eight non-overlapping fields examined.

This value should be interpreted as the apparent area fraction of the dark regions revealed by optical microscopy after chemical etching, rather than as the isolated fraction

of a specific intermetallic phase. These dark regions include interdendritic segregated areas and secondary constituents formed during solidification. Their association with intermetallic-rich regions was supported by SEM/EDS analyses, which revealed localized Al- and Mn-enrichment. XRD independently confirmed β -Mg₁₇Al₁₂ but did not reliably resolve the Al–Mn-rich constituents.

The SEM/EDS maps acquired from ten representative fields provided chemical support for the interpretation of the optical micrographs. The Mg maps showed a predominant Mg-rich matrix, while the Al and Mn maps revealed localized enriched regions associated with the darker interdendritic/secondary constituents observed by optical microscopy. These results indicate that the dark regions revealed by chemical etching are chemically heterogeneous and are associated with Al- and Mn-enriched constituents formed during solidification.

The heterogeneous distribution of these interdendritic/secondary regions is associated with the solidification behavior of Mg–Al–Mn alloys. During the final stages of solidification, solute redistribution promotes the formation of Al-rich and Mn-rich constituents in interdendritic regions and along grain boundaries, producing a microstructure composed of a predominant α -Mg matrix and a smaller fraction of secondary phases.

Therefore, the optical area fraction provides a broader estimate of the microstructural heterogeneity of the cast AM50 alloy. This heterogeneity is relevant to the mechanical behavior, since the darker interdendritic/secondary regions may contribute to local hardness variations, strain partitioning, stress concentration, and premature crack initiation during tensile loading. The coexistence of a predominant α -Mg matrix with localized secondary constituents contributes to the mechanical contrast within the microstructure and helps explain the scatter observed in hardness and tensile properties.

3.3. Local and Global Hardness Response

Table 1 presents the average hardness values obtained for the AM50 alloy. The results demonstrate a mechanical response directly associated with the microstructural heterogeneity observed in the OM and SEM analyses.

Table 1. Hardness obtained for the AM50 alloy.

Hardness	Average	Standard Deviation
Matrix, HV	39	5
Constituent-rich region, HV	119	12
Matrix, HK	61	10
Constituent-rich region, HK	164	36
HRF	33	4
HB	54	3

The Vickers and Knoop microhardness tests revealed significant differences between the α -Mg matrix and the constituent-rich regions. The highest local hardness values were measured in regions containing Al- and Mn-rich precipitates, particularly in areas associated with β -Mg₁₇Al₁₂, indicating that constituent-rich regions act as locally hardened regions.

SEM images of the indentation sites were not available for this analysis; therefore, the local hardness results were interpreted based on the corresponding microstructural features identified by optical microscopy and SEM/EDS.

This difference in local hardness is directly related to the restriction of dislocation mobility promoted by the intermetallic phases present within these regions. While the α -Mg matrix, with an HCP structure, presents a greater capacity for plastic deformation,

the secondary phases exhibit high stiffness and reduced deformability, promoting a non-uniform distribution of strain across the microstructure.

The dispersion observed in the microhardness results also reinforces the heterogeneous distribution of secondary phases in the cast alloy. Regions containing a higher concentration of precipitates showed higher local hardness values, whereas regions predominantly composed of the α -Mg matrix exhibited lower values. This behavior is consistent with the microstructural results obtained by SEM/EDS.

The Brinell and Rockwell F hardness results represent the global mechanical response of the alloy, simultaneously integrating the contribution of the matrix and the intermetallic phases. The value of 54.2 HB is in reasonable agreement with the results reported by [10,26], while the low dispersion of the values indicates a relatively uniform macroscopic response, despite the heterogeneity at the microstructural scale.

The mechanical incompatibility between the ductile matrix and the rigid intermetallic phases directly influences the deformation mechanisms of the alloy. During mechanical loading, the α -Mg matrix is expected to accommodate strain primarily through dislocation glide and twinning mechanisms, which are characteristic of magnesium alloys due to the limited number of active slip systems associated with the HCP crystal structure [27,28]. In contrast, the intermetallic phases, due to their higher stiffness and lower deformability, tend to exhibit a more restricted mechanical response under applied stress [24].

Consequently, the mechanical incompatibility between the α -Mg matrix and the intermetallic phases may favor localized stress concentration at the matrix/intermetallic interfaces, potentially contributing to dislocation accumulation and increasing the susceptibility to crack initiation and interfacial decohesion [29]. Furthermore, the presence of these rigid phases may intensify strain partitioning between different microstructural regions, suggesting a tendency toward non-uniform deformation at the microscopic scale [30].

The influence of constituent-rich regions on local hardness has also been reported under different heat treatment conditions. According to [31], as-cast samples exhibited higher hardness than those treated at 410 °C and lower hardness than those treated at 380 °C, reaching approximately 61 HV. In aged samples, the hardness increased from approximately 60 HV at 150 °C to about 63 HV at 200 °C, followed by a subsequent decrease to values close to the initial condition.

Thus, the results indicate that the microstructural heterogeneity of the AM50 alloy contributes significantly to its mechanical heterogeneity. The interaction between the more ductile α -Mg matrix and the harder intermetallic-rich regions influences the deformation mechanisms, stress distribution, and damage evolution, contributing to the overall mechanical behavior of the alloy.

3.4. Tensile Behavior and Mechanical Scatter

The tensile behavior of the cast AM50 alloy was evaluated through uniaxial tensile tests. The engineering stress-strain curves obtained from tensile testing are presented in Figure 6, allowing us to note the influence of microstructural dissimilarity on the evaluation of the mechanical response, tensile strength, and deformation capacity of the alloy.

The specimens exhibited an average ultimate tensile strength of 129.42 ± 17.63 MPa and an average maximum strain of $0.80 \pm 0.98\%$. The coefficients of variation were 13.62% for ultimate tensile strength and 122.43% for maximum strain, indicating a considerably larger scatter in ductility than in strength. Only six specimens were tested; therefore, the mean, standard deviation, and coefficients of variation are considered descriptive statistics. The CV was 13.6% for UTS and 122.4% for recorded fracture strain, meaning that the relative scatter in strain was approximately nine times greater than that observed for UTS.

Accordingly, the mean fracture strain should not be interpreted as a robust population value, and no inferential statistical claim is made.

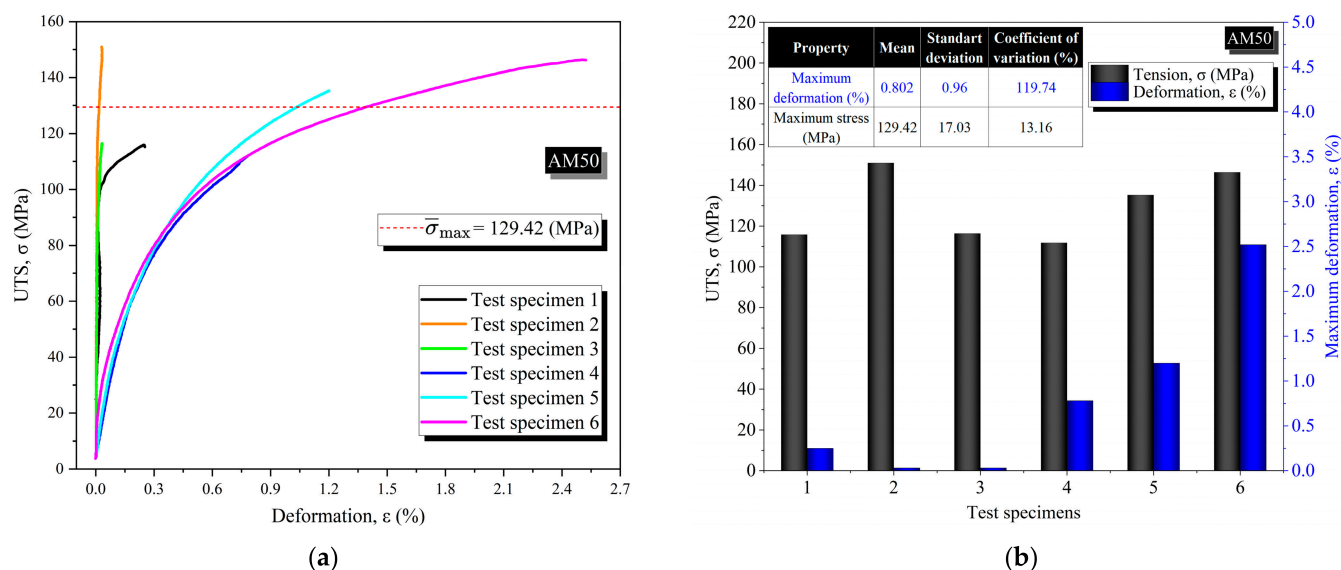


Figure 6. Tensile test: (a) stress-strain curves of the test and (b) comparison of the performance of each test specimen. The initial slopes are presented only as recorded experimental responses and were not used to determine Young's modulus.

The initial slopes of the recorded engineering stress-strain curves varied substantially among specimens. These slopes are not interpreted as Young's modulus because the test configuration was not intended for precision modulus determination and may include contributions from extensometer seating, specimen alignment, machine/grip compliance, and the limited strain interval before premature fracture. Accordingly, the initial slopes are excluded from the microstructure–property correlation.

The results showed substantial scatter among the specimens, particularly in the maximum strain to fracture, while the variation in ultimate tensile strength was comparatively smaller. Some specimens exhibited premature rupture with limited plastic deformation, whereas others displayed comparatively greater deformation before fracture. This difference in deformation capacity indicates that local conditions within the cast material may have played an important role in determining the fracture response.

This behavior can be discussed in relation to the heterogeneous distribution of intermetallic phases observed in the microstructural analyses. Regions containing higher concentrations of β -Mg₁₇Al₁₂ and Al–Mn intermetallic particles may provide localized strengthening; however, these constituents can also restrict plastic accommodation within the surrounding α -Mg matrix. Consequently, interfaces between the matrix and secondary phases may act as regions of local mechanical contrast and potential stress concentration during uniaxial loading.

Furthermore, the SEM images suggest the presence of small discontinuities and possible surface cracks associated with brittle constituents, reinforcing the possibility that casting- or microstructure-related discontinuities contributed to premature fracture in certain tensile specimens. One study shows that microhardness analyses performed in different regions of the AM50 alloy demonstrated significant differences between the matrix and intermetallics, indicating a high local mechanical contrast [32]. This property gradient may contribute to localized stress and strain partitioning during deformation. However, the present observations do not allow the contribution of intermetallic morphology to be quantitatively separated from that of porosity, inclusions, segregation, or other casting-related defects.

The mechanical behavior of magnesium alloys is strongly influenced by their hexagonal close-packed (HCP) crystal structure, which limits the number of independent slip systems active at room temperature [28]. According to the literature, tensile deformation in Mg alloys is commonly associated with the activation of basal slip systems and deformation twinning, particularly in alloys exhibiting heterogeneous microstructures [16,25].

In the present study, the significant scatter observed in the stress-strain curves suggests that local microstructural variations, including the heterogeneous distribution of intermetallic phases and possible stress concentrators, may have influenced the deformation behavior and contributed to premature fracture in some specimens. The combination of a limited field-specific area fraction of Al/Mn-enriched constituents, pronounced local hardness contrast, and exceptionally high strain scatter indicates that intermetallic content alone is insufficient to explain the tensile variability. Therefore, the experimental results support a multi-factor interpretation rather than a single-phase causal model.

The differences observed among the specimens may also be related to the presence of microstructural segregations, porosity, and defects originating from the solidification and machining processes. Although the tests were performed in accordance with the ASTM E8/E8M-24 standard, the high sensitivity of magnesium alloys to local heterogeneities can result in significant scatter in mechanical results. AM50-specific quantitative fractography reported in the literature indicates that tensile ductility can correlate more strongly with porosity present on the corresponding fracture surface and with localized pore clusters than with average bulk porosity. The fracture path may preferentially cross regions containing clustered gas and shrinkage pores. Because fracture-surface porosity, maximum pore size, and pore distribution were not quantitatively characterized in the present study, the contribution of casting defects cannot be separated quantitatively from that of intermetallic morphology, inclusions, or segregation.

The average UTS of 129.42 ± 17.63 MPa is below values reported for optimized low-pressure die-cast AM50 and thin-section HPDC AM50 alloys, but published values vary substantially with casting route, section thickness, and sampling condition [16,33]. For example, UTS values have been reported to decrease from approximately 240 MPa in 2 mm HPDC sections to approximately 112 MPa in 10 mm sections, while a recent gravity-cast AM50 reference alloy exhibited a UTS near 110 MPa [12]. Accordingly, the present UTS is not treated as anomalous in isolation; rather, it is interpreted as a response specific to the investigated ingot and sampling condition.

However, previous studies on cast AM50 magnesium alloys containing inclusions and microstructural heterogeneities have also reported limited ductility associated with mixed brittle-ductile fracture behavior [34]. One study reported ultimate tensile strength values around 185 MPa combined with deformation close to 1.5% in directly cast AM50 alloys containing oxide-rich and Fe-rich inclusions [34]. According to that study, the presence of inclusions, brittle intermetallic phases, and local microstructural discontinuities promoted stress concentration and premature crack initiation. These observations provide a relevant comparison with the heterogeneous tensile response observed in the present study, although the specific contribution of each type of defect cannot be established quantitatively from the present dataset.

Overall, the results demonstrate that the mechanical behavior of the AM50 alloy is influenced by the interaction between the α -Mg matrix, intermetallic phases, and casting-related heterogeneities. The coexistence of relatively ductile α -Mg regions and harder secondary constituents can contribute to a heterogeneous mechanical response, while porosity, inclusions, segregation, and other local discontinuities may additionally affect the onset and propagation of fracture.

The fractographic observations provide further evidence of this heterogeneous fracture response. The coexistence of dimples, tear ridges, and locally brittle or rough regions establishes a mixed fracture morphology, involving both localized plastic deformation and regions with more brittle characteristics. However, these features alone do not identify the specific crack-initiation defect unless the initiation site is located and subsequently characterized chemically and morphologically. Therefore, the present fractographic observations support the occurrence of mixed fracture mechanisms but do not allow a definitive assignment of crack initiation to intermetallic particles, inclusions, porosity, or segregation.

Thus, the correlation between the microstructural analyses and tensile test results indicates that the distribution, morphology, and local mechanical contrast associated with secondary phases contribute to the mechanical response of the AM50 alloy, but they should not be considered the sole explanation for the observed tensile variability. The combination of microstructural heterogeneity and possible casting-related discontinuities provides a more consistent interpretation of the pronounced scatter in fracture strain observed among the six tested specimens.

3.5. Fracture Surface Analysis

The fractographic micrographs of specimens 1, 2, and 3 are presented in Figure 7, revealing fracture patterns characteristic of both ductile and brittle failure modes, strongly influenced by the microstructural heterogeneity of the AM50 alloy. On the lower fracture surfaces, pronounced roughness and irregular dimples were observed, indicating the nucleation of microcracks at stress concentration sites, probably associated with interdendritic regions enriched in intermetallic phases observed in the polished-section SEM/EDS analyses. In addition, the presence of tear ridges was identified in several regions of the fracture surface, indicating localized plastic deformation during crack propagation and suggesting the coexistence of ductile deformation mechanisms with brittle fracture features. Such features suggest that microstructural heterogeneities may have contributed to localized damage accumulation and crack propagation during tensile loading. This observation is consistent with previous studies reporting the fracture behavior of cast Mg-Al-Mn alloys, in which secondary phases act as preferential sites for crack initiation and propagation [23,24].

When the fractographic features are compared with the tensile response of each specimen, differences in the relative occurrence of ductile and brittle features are apparent. Specimens with lower recorded deformation showed a greater relative presence of rough fracture regions and localized brittle features, whereas specimens with comparatively higher deformation exhibited more dimples and tear ridges. These observations are consistent with different degrees of local plastic accommodation, but they do not identify the specific crack-initiation defect or establish a unique causal relation with intermetallic-rich regions.

On the upper fracture surfaces, more uniform dimples and lower surface roughness were observed, suggesting that fracture occurred with a lower degree of localized plastic deformation. Tear ridges were also identified between adjacent fracture regions, indicating localized plastic accommodation during crack propagation. Although 1, 2, and 3 exhibited similar overall fracture morphologies, differences in the relative extent of ductile and brittle regions were observed. Therefore, while the dominant fracture mechanisms appear to be similar among the specimens, local microstructural variations may have influenced the fracture path and contributed to the scatter observed in the tensile properties.

Secondary constituents identified by SEM/EDS and XRD contribute to local mechanical contrast because constituent-rich regions are harder and less deformable than the surrounding α -Mg matrix. Such contrast may favor local stress and strain partitioning. Nevertheless, the present fractographic observations do not demonstrate that crack initia-

tion was controlled specifically by intermetallic particles. Porosity, oxide films, inclusions, segregation, and other casting-related discontinuities remain plausible interacting contributors and were not quantitatively separated in this study. Figure 8 presents the fractographic analysis of specimens 4, 5, and 6, which exhibited greater deformation before fracture and a more pronounced contribution of ductile features.

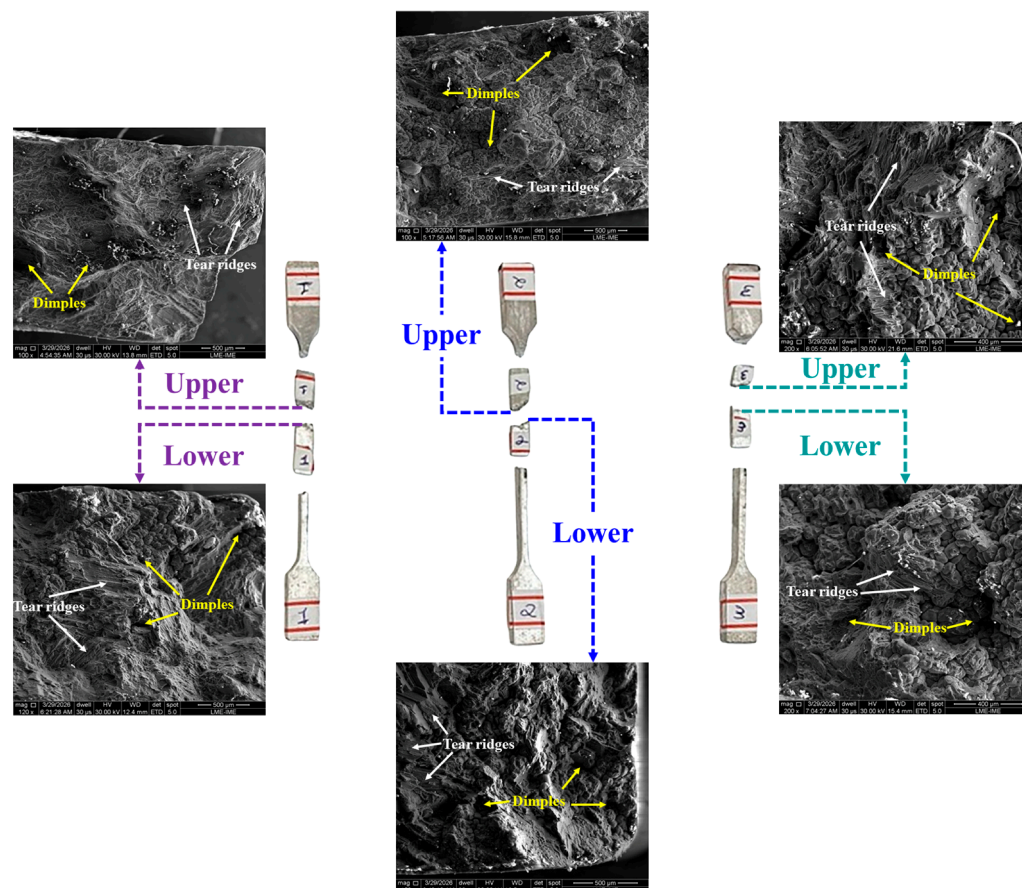


Figure 7. Fractographic analysis of tensile-tested specimens 1, 2, and 3. The yellow arrows indicate dimples, while the white arrows identify tear ridges, highlighting the mixed ductile–brittle fracture morphology observed on the analyzed surfaces.

In contrast to specimens 1, 2, and 3, specimens 4, 5, and 6 exhibited greater deformation before fracture and showed a comparatively stronger contribution of localized plastic deformation on their fracture surfaces. The fractographic analysis of specimen 4 revealed a mixed fracture morphology with a pronounced ductile contribution, particularly at 400 $\times$ and 800 $\times$ magnification, where rough and tortuous regions, ligaments, tear ridges, and depressions or cavities associated with localized deformation were observed. These features indicate that the material underwent greater plastic accommodation during crack propagation before final fracture. Although some relatively flat regions remained, their lower predominance compared with the fracture surfaces of specimens 1–3 is consistent with the greater deformation capacity observed for specimen 4.

Specimens 5 and 6 also exhibited mixed fracture morphologies, although with a high degree of local heterogeneity. In specimen 5, the images at 200 $\times$ and 400 $\times$ showed the coexistence of relatively large faceted regions with rougher and apparently deformed areas, indicating the simultaneous occurrence of different fracture mechanisms. In specimen 6, the contribution of localized plastic deformation was more evident, particularly at 400 $\times$ and 800 $\times$, where numerous tear ridges, ligaments, and depressions compatible with localized microvoid nucleation and coalescence were observed. Therefore, although both specimens

retained regions with faceted morphology, the presence of more extensive rough and plastically deformed regions is consistent with their comparatively greater deformation before fracture.

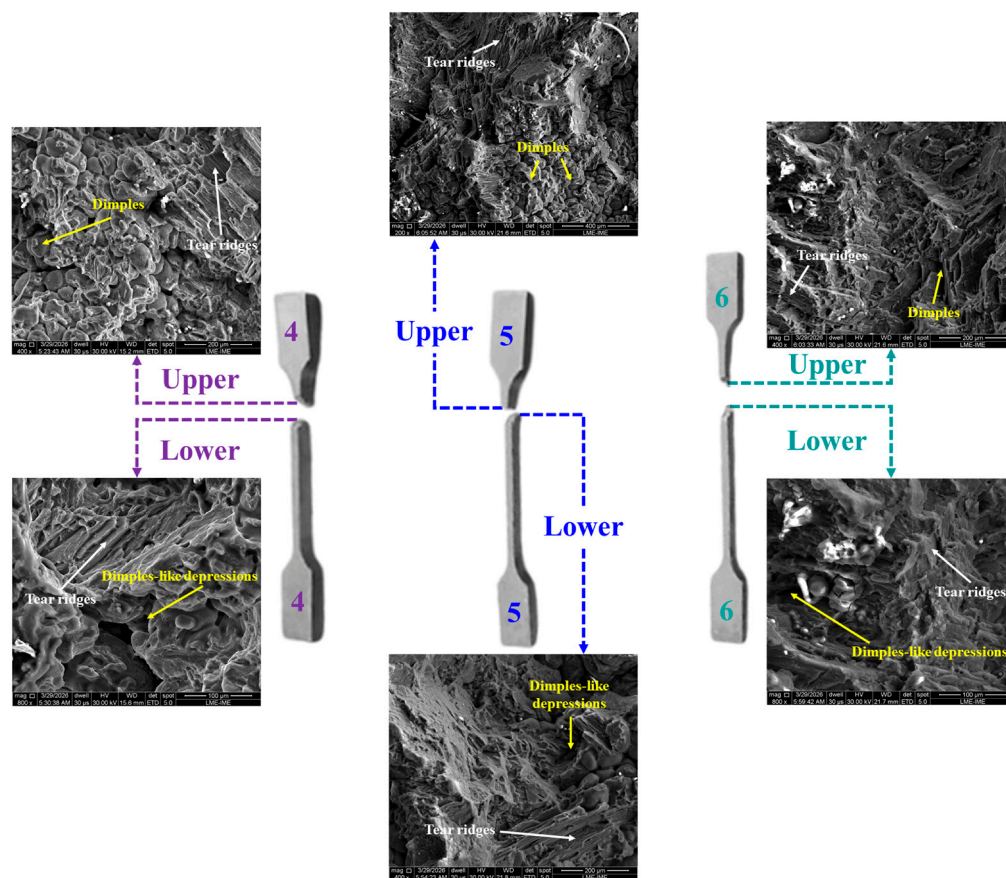


Figure 8. Fractographic analysis of tensile-tested specimens 4, 5, and 6, showing mixed fracture morphologies with a more pronounced contribution of localized plastic deformation. Yellow arrows indicate dimples or dimple-like depressions, while white arrows indicate tear ridges.

Overall, the fractographic features of specimens 4, 5, and 6 support the interpretation that their higher deformation was associated with a greater capacity for localized plastic accommodation before final fracture. Nevertheless, the fracture surfaces cannot be classified as purely ductile, since faceted regions and features compatible with quasi-cleavage remained present, particularly in specimens 5 and 6. Thus, these specimens can be characterized as exhibiting mixed fracture behavior, with a more pronounced ductile contribution than that observed in specimens 1–3. The coexistence of rough, plastically deformed regions with faceted areas indicates that the local microstructural condition influenced the balance between plastic deformation and crack propagation mechanisms, helping to explain the higher strain-to-fracture values observed for specimens 4–6.

3.6. Microstructure–Property Correlation

The results indicate that the as-cast AM50 alloy contains a heterogeneous microstructural arrangement composed of a predominant α -Mg matrix and localized Al- and Mn-enriched secondary constituents. Their irregular morphology and preferential interdendritic distribution are relevant to the local mechanical response, as demonstrated by the pronounced hardness contrast.

The softer α -Mg matrix provides greater capacity for plastic accommodation, whereas harder constituent-rich regions restrict local deformation. This mismatch can promote

strain partitioning and local stress concentration at matrix/constituent interfaces. These observations support a micromechanical role for secondary constituents without implying that they are the sole determinant of the bulk tensile response.

The much greater relative scatter in recorded fracture strain than in UTS indicates that fracture-sensitive behavior was strongly affected by local conditions. Because the tensile specimens were extracted from comparable upper regions of different commercial ingots, the observed variability may reflect the combined effects of constituent morphology, local segregation, casting-related discontinuities, sampling history, and between-ingot variability.

The fractographic observations are consistent with this multi-factor interpretation. The coexistence of dimples and tear ridges with rougher and locally brittle regions confirms a mixed ductile–brittle fracture morphology, but the analyzed surfaces do not permit unequivocal identification of the dominant crack-initiation site. Thus, the fractography complements the hardness and microstructural results while also defining the limits of the present causal interpretation.

Overall, the intermetallic-rich constituents contribute to local mechanical heterogeneity, whereas the macroscopic tensile variability cannot be attributed to intermetallic content or morphology alone. A more complete quantitative correlation would require position-resolved defect characterization, fracture-surface porosity measurements, and a larger tensile dataset.

4. Conclusions

In this study, the microstructure, local hardness response, tensile behavior, and fracture characteristics of commercially produced as-cast AM50 magnesium alloy were evaluated. The main conclusions are:

1. The microstructure was dominated by α -Mg with Al-enriched interdendritic constituents consistent with β -Mg₁₇Al₁₂ and Al–Mn-rich particles. XRD confirmed α -Mg and β -Mg₁₇Al₁₂; the Al–Mn-rich constituents were supported by SEM/EDS but could not be reliably indexed by XRD.
2. Optical image analysis of eight non-overlapping fields yielded an apparent area fraction of $27.6 \pm 1.1\%$ for dark interdendritic/secondary regions. This value represents the combined etched microstructural constituents and should not be interpreted as the fraction of an individual intermetallic phase.
3. A pronounced local hardness contrast was measured between the α -Mg matrix and constituent-rich regions (39.1 ± 5.3 HV and 119 ± 12 HV, respectively), demonstrating substantial micromechanical heterogeneity.
4. Six tensile specimens extracted from comparable upper regions of different commercial ingots exhibited an average UTS of 129.42 ± 17.63 MPa. Recorded fracture strain showed much greater relative scatter than UTS (CV $\approx 122\%$ versus $\approx 13.6\%$); given the limited sample size, these statistics are descriptive.
5. The measured UTS lies within the broad range reported for AM50 processed under different casting and section-thickness conditions. Accordingly, it should be interpreted in relation to casting route, local solidification history, sampling position, and possible between-ingot variability rather than classified as intrinsically anomalous.
6. Fractography revealed mixed ductile–brittle features, but the crack-initiation sites were not unequivocally identified. Intermetallic-rich regions contribute to local mechanical contrast, while porosity, oxide films, inclusions, segregation, and other casting-related discontinuities remain plausible interacting contributors.
7. The principal contribution of the study is the demonstration that pronounced local hardness contrast coexists with tensile variability that cannot be explained by

intermetallic constituents alone. Quantitative separation of the controlling mechanisms will require position-resolved defect characterization, fracture-surface porosity measurements, and a larger tensile dataset.

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Abbreviations

FEG	Field-Emission Gun
HB	Hardness Brinell
HCP	Hexagonal Close-Packed
HK	Knoop Microhardness
HRF	Rockwell F Hardness
HV	Vickers Microhardness
OM	Optical Microscopy
SEM	Scanning Electron Microscopy
XRD	X-ray Diffraction

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