



Effect of Heat Treatment on the Fatigue Characterization of an Automotive Component Produced by HPDC Using a Recycled AlSi10MnMg Alloy

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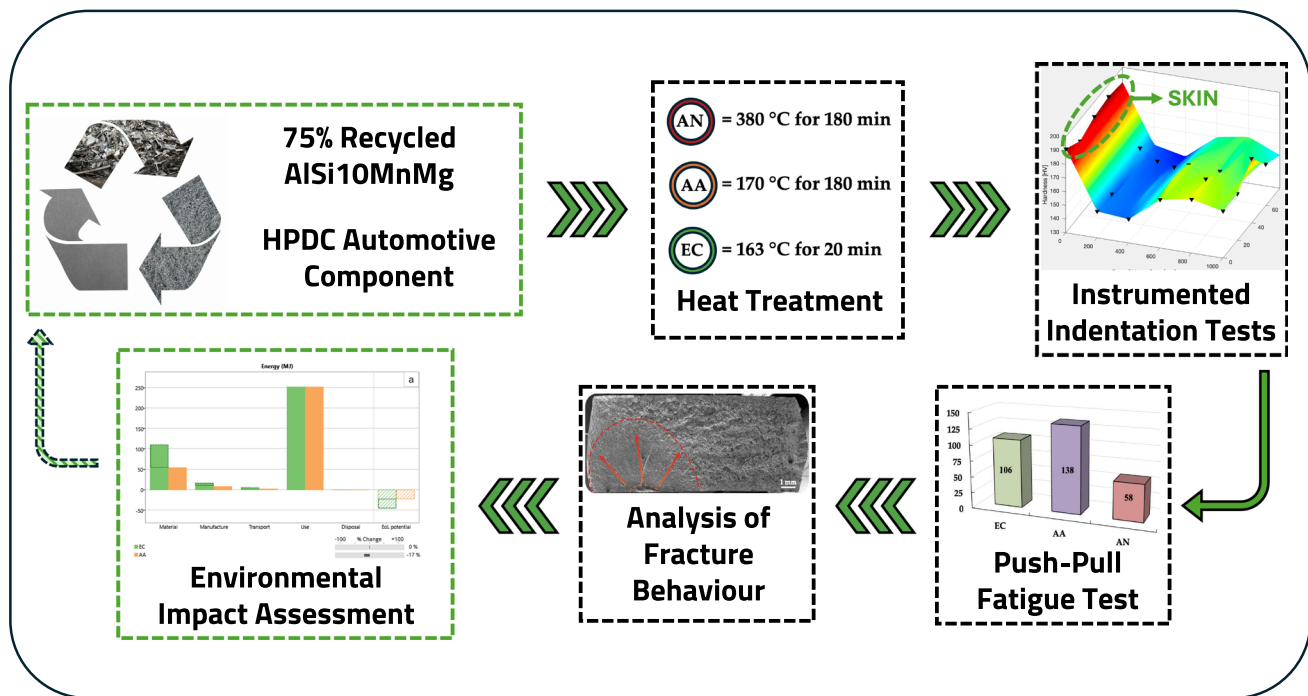
Al–Si–Mg alloys are increasingly employed in HPDC automotive structural components due to their lightweight potential and recyclability. In this context, the use of recycled AlSi10MnMg alloys represents a promising strategy to reduce the environmental footprint of aluminum components. However, the fatigue performance of these alloys, particularly when subjected to different heat treatments, remains insufficiently understood. This study investigates the fatigue behavior and fracture mechanisms of a 75 pct recycled AlSi10MnMg alloy, using specimens directly extracted from an industrial HPDC automotive component. Specimens retaining the original as-cast surface were subjected to three different heat treatment conditions: E-coating simulation (EC), artificial aging (AA), and annealing (AN). Microstructural characterization, nanoindentation mapping, and high-cycle fatigue tests were combined with fractographic analysis to correlate microstructural features with fatigue crack nucleation and propagation mechanisms. Results revealed a heterogeneous cross-sectional microstructure, with a discontinuous eutectic-rich skin layer over a coarser dendritic core, generating a hardness gradient that significantly influenced fatigue behavior. Among the investigated conditions, AA provided the highest fatigue strength (138 ± 4 MPa), while EC and AN showed lower performance. Fractographic observations revealed that fatigue crack nucleation was associated with the local discontinuity of the skin layer, whereas crack propagation and overall fatigue resistance were primarily governed by the hardness distribution induced by heat treatment. Finally, the environmental impact assessment focused on the effect of the AA and EC heat treatments further indicated that AA, despite its additional processing energy, reduces the overall life-cycle environmental burden by extending component lifetime.

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I. INTRODUCTION

ALUMINUM alloys have gained increasing attention in the automotive industry due to their high strength-to-weight ratio, corrosion resistance, and recyclability.^[1] Beyond traditional applications in powertrain and transmission systems, the next generation of vehicles will increasingly rely on large structural castings for body-in-white and dedicated components for electric vehicles (EVs), for example, shock towers and battery cases, thereby reinforcing aluminum as a key enabler of lightweight and sustainable design.^[2]

To offset the additional mass introduced by the battery pack in EVs, lightweight materials such as aluminum–silicon alloys are widely employed, offering a density roughly one-third of steel. In particular, Al–Si–Mg alloys are increasingly used for large castings, such as battery and e-motors/e-drives housing, due to their excellent thermal conductivity,^[3] which enhances cooling efficiency during operation. Similarly, in traditional internal combustion engine (ICE) vehicle castings, these alloys are extensively employed for engine blocks, transmission cases, and cylinder heads^[4] where structural integrity under static and dynamic loads is essential.^[5]

High-pressure die casting (HPDC) is a manufacturing technology that can support the transition to a more sustainable mobility, while also meeting the demanding mechanical and thermal requirements of contemporary

EVs architectures.^[6] This latter method is extremely efficient for mass manufacturing of complex-shaped thin wall components, with fast production speed, high dimensional accuracy, good surface finish, and competitive costs.^[7,8]

Hypoeutectic Al–Si alloys with near-eutectic compositions, such as primary AlSi10Mg, have been increasingly adopted in recent years for automotive structural HPDC castings, where enhanced mechanical performance and damage tolerance are required.^[9] Traditionally, HPDC production has relied on recycled Al–Si alloys, extensively used for non-structural and semi-structural components due to their excellent castability and cost efficiency. More recently, the demand for large, lightweight structural components has promoted the use of primary Al–Si–Mg alloys, while growing environmental concerns have driven interest in recycled Al–Si–Mg alloys,^[10] as a more sustainable alternative to primary alloys.^[11]

Nonetheless, using recovered alloys introduces new design and compositional challenges, because recycled alloys frequently show greater amounts of Fe as a result of the recycling process, which can negatively affect mechanical properties.^[12,13] Although Fe decreases die soldering phenomenon,^[14] thereby promoting the use of recycled alloys in HPDC, these are normally not applied for safety-critical or high-ductility components due to the formation of brittle β -Al₅FeSi intermetallic compounds, which compromise their mechanical

performance.^[13] Moreover, these intermetallic compounds obstruct the flow of interdendritic liquid during the final stages of solidification, thereby hindering effective feeding and promoting the formation of shrinkage-related voids. The detrimental effect of Fe on porosity formation is particularly evident in regions of castings characterized by marginal solidification conditions, such as poorly fed hot spots where the risk of shrinkage porosity is amplified.^[15]

To mitigate the detrimental effects associated with Fe-based intermetallic compounds, industrial practice generally aims to limit the Fe content to low levels, while introducing suitable alloying elements to control phase morphology. In this context, Mn is commonly added to promote the formation of more compact and less harmful $\alpha\text{-Al}_{15}(\text{Fe},\text{Mn})_3\text{Si}_2$ intermetallics, which exhibit a reduced negative impact on mechanical properties.^[16] In addition, Mn plays a key role in improving die soldering resistance and reducing metal–die interaction phenomena during HPDC, thereby enhancing process stability and die life, even when Fe contents are maintained at relatively low levels. For these reasons, the most widely used recycled alloy for HPDC components is the AlSi10MnMg alloy.

Another way to increase mechanical properties in Al–Si–Mg alloys is through heat treatment, particularly solution heat treatment followed by quenching and artificial aging (T6). This treatment is commonly applied to enhance the tensile strength and ductility in Al–Si–Mg alloys produced by both sand and permanent mold casting. When applied to HPDC components, however, the solution phase could lead to a growth of pores, as depicted by Hu *et al.*^[17] leading to blistering phenomenon and distortion of high-dimension castings. For this reason, direct aging (T5) is generally preferred, as it maximizes strength while preserving dimensional stability.

More recently, from both an industrial and sustainability perspective, the need for dedicated heat treatments is being re-evaluated, since HPDC components are often subjected to post-processing operations that already involve significant thermal exposure. In particular, surface finishing processes such as electrophoretic deposition (E-coating)^[18,19] are typically applied at temperatures above 150 °C. Such thermal exposures can effectively act as a low-temperature aging treatment, promoting precipitation phenomena similar to those induced by a conventional T5 condition. As a result, E-coating has been increasingly considered as a potential functional substitute for a dedicated artificial aging treatment, enabling mechanical strengthening while reducing energy consumption and processing steps. However, when ductility or conductivity are the main component requirements, the application of an annealing treatment^[20] may represent a valuable option for HPDC alloys, as it reduces hardness by softening the aluminum matrix and relieves residual stresses introduced during the casting process.^[21]

Despite the high static mechanical strength typically achieved in HPDC alloys, owing to rapid cooling rates and refined microstructures,^[22] the presence of casting-related defects such as porosity, oxide films, and

cold joints^[23] remains a major limitation for cyclic fatigue resistance, as also reported by Lattanzi *et al.*^[7] In addition, as reported by Dalai *et al.*,^[24] some HPDC castings may also exhibit an inhomogeneous skin layer, where the surface is not continuous due to the complex fluid flow patterns inside the die cavity, leading to spatial variations in heat transfer and cooling rate during solidification.

The fatigue behavior of primary AlSi10Mg alloys has been investigated in a limited number of studies, mainly focusing on specimens produced through processes such as laser powder bed fusion.^[25,26] However, to the best knowledge of the authors, no fatigue data have been reported in the literature for the recycled AlSi10MnMg alloy used in HPDC, despite its recent growing use in automotive components.

The aim of this work was to investigate the fatigue fracture mechanisms in samples directly extracted from an industrial HPDC automotive component made of a recycled AlSi10MnMg alloy. To achieve this goal, a comprehensive microstructural, mechanical, and fractographic characterization was carried out. Mechanical testing included fatigue and nanoindentation tests carried out on specimens retaining the original casting features, without any material removal.

Since fracture mechanisms are strongly influenced by the microstructure and its evolution during heat treatments, the effects of three different conditions were investigated: E-coating simulation (EC), artificial aging (AA), and annealing (AN).

Finally, an environmental impact assessment focused on the effect of the AA and EC heat treatments was conducted to evaluate the implications of different heat treatments on the life cycle of a recycled automotive casting, including simulations of the use and transportation phases.

II. MATERIALS AND METHODS

A. Material

The samples studied in this research were extracted from an automotive component produced *via* HPDC using a 75 pct recycled (HR) AlSi10MnMg alloy. The component was manufactured according to automotive specifications for a premium car manufacturer and had wall thicknesses ranging from approximately 3 to 4 mm, with an average thickness of about 3.8 mm and a maximum overall dimension of 1200 mm. The chemical composition of the alloy was determined by glow discharge optical emission spectroscopy (GD-OES, GDA-650, Spectrumba Analytik GmbH) and is reported in Table I. The composition complies with the standard specifications^[27] for AlSi10MnMg alloys. The alloy was supplied by Raffmetal S.p.A. as secondary ingots with 75 pct recycled content, obtained by remelting scrap mainly derived from end-of-life products mixed with 25 pct primary aluminum. Ingots were melted in a reverberatory furnace, with the charge including up to 40 pct internal foundry returns.

Upon extraction from the die, the castings were rapidly water-quenched at the end of solidification to enable trimming; this condition is referred to as the as-produced (AP) state. Three different heat treatments (HTs) were investigated: E-coating simulation (EC), artificial aging (AA), and annealing (AN). For the EC, the samples in the AP condition were placed in a cold furnace then heated to the defined temperature, whereas for the AA and AN, the samples were inserted into the furnace only after the target temperature had been reached.

The parameters selected for the heat treatments were chosen based on previous studies on the alloy conducted by the authors^[28,29] and are summarized in Figure 1.

Table II summarizes the tensile properties reported in a previous study^[29] for the heat treatments shown in Figure 1. The mechanical strength is represented by the yield strength (YS) and ultimate tensile strength (UTS), while ductility is expressed in terms of elongation to failure (e pct).

The reported tensile properties allow a preliminary comparison of the mechanical response of the alloy under the investigated heat treatments, providing a reference framework for interpreting the fatigue behavior, which is the primary focus of the present study.

B. Microstructural Characterization

Metallographic samples were prepared following standard procedures for microstructural analysis. Specimens were embedded in conductive resin, ground with emery papers up to P2500 grit, and subsequently polished using polycrystalline diamond suspensions down to 1 μm , in accordance with ASTM E3-11(2017) standard.^[30] Metallographic etching was performed with Keller’s reagent (95 mL H₂O, 2.5 mL HNO₃, 1.5 mL HCl, 1.0 mL HF) for 10 s, following ASTM E407-07(2015)e1 standard.^[31] Microstructural characterization was carried out using both optical microscopy (AXIOSCOPE 7, Zeiss) and field-emission gun scanning electron microscopy (SEM-FEG, Tescan Mira3) equipped with energy-dispersive X-ray spectroscopy (EDS). The investigation focused on phase distribution, casting defects (*e.g.*, porosity, inclusions), skin layer and Fe-based intermetallic compounds. Quantitative image analysis was conducted using ImageJ software to evaluate key microstructural features, including secondary dendrite arm spacing (SDAS), defect morphology, and evaluation of skin layer thickness. The extracted data provided essential metrics to correlate microstructural characteristics with the mechanical behavior of the alloy under cyclic loads. Statistical analysis was performed to ensure the reproducibility and significance of the results.

C. Instrumented Indentation Tests

Indentation hardness (HIT) was evaluated by Instrumented Indentation Testing using a high-resolution nanoindentation system (PB1000, Nanovea) equipped with a Berkovich diamond tip. Before tests, the Berkovich indenter underwent calibration with a fused

Table I. Chemical Composition (Weight Percent) of the HR AlSi10MnMg Measured by GD-OES

Alloy Designation	Si	Mn	Mg	Fe	Cu	Zn	Ti	Sr	Al
75 pct Recycled AlSi10MnMg	10.35 ± 0.04	0.49 ± 0.01	0.45 ± 0.01	0.13 ± 0.01	0.06 ± 0.01	0.07 ± 0.01	0.07 ± 0.01	0.018 ± 0.01	bal

silica block, in compliance with BS EN ISO 14577-1:2015.^[32] Indentations were performed in depth-control mode with maximum penetration depth set as 0.70 μm to get a lower projected area compared to skin layer thickness. A loading/unloading rate of 30 mN/min was applied for each test. The Oliver–Pharr methodology,^[33] extensively validated for aluminum alloys,^[34] was used to calculate HIT from indentation curve. For each condition, 30 indentations were arranged in matrices consisting of 5 vertical impressions spaced 0.02 mm apart, repeated in six series 0.2 mm far from each other, as shown in Figure 2.

The depicted layout extended from the skin layer to the core of the material, enabling a systematic mapping of the mechanical response across the sections (*i.e.*, skin layer, transition zone and core) while preventing interference between adjacent plastic zones.

D. Fatigue Tests

Fatigue specimens were extracted from flat regions of the casting with an available area of approximately 200 $\times$ 350 mm. For each condition, fatigue specimens were machined *via* electrical discharge machining

Table II. Mechanical Properties of the HR AlSi10MnMg Alloy in the EC, AA, and AN Condition in Terms of Yield Strength (YS), Ultimate Tensile Strength (UTS), and Elongation to Failure (e pct)

	YS	UTS	e Pct
	[MPa]	[MPa]	[Pct]
EC	164	301	7
AA	239	344	6
AN	108	195	16

(EDM) to reach the geometry shown in Figure 3(a), according to the ASTM E466-21 standard.^[35] This process was chosen to preserve the original casting features of the HPDC component, including the as-cast surface skin and the full casting thickness. The specimens were therefore tested in a condition representative of the industrial casting surface, where the skin layer may be locally developed with different thickness and continuity depending on the local solidification conditions. Only after machining, the gauge length surfaces were lightly ground and polished using P2500 grit SiC paper, exclusively to remove the EDM heat-affected layer ($\approx$ 0.1 mm per side) and to achieve the target surface roughness, without altering the underlying microstructure.

All the fatigue tests were performed at room temperature on a resonant testing machine at 110 Hz. The system consists of a set of resonance springs and the specimen, which together form an oscillating unit tuned to the natural frequency of the frame. The electromagnetic actuator maintains the oscillation, while a mechanical preloading device introduces the mean load required to achieve the selected stress ratio. The specimen is clamped at the center of the machine, where the oscillatory forces are transmitted, as shown in the experimental layout reported in Figure 3(b).

A stress ratio of $R = 0.2$, corresponding to a tension-tension cycle with $\sigma_{\min} = \sigma_{\max} \cdot 0.2$, was applied to generate a severe testing condition consisting exclusively of tensile loads, representing a compromise between the load capacity of the machine and the choice of $R = 0.1$, which is often adopted in automotive component testing. The applied load was measured and controlled through an integrated 20 kN load cell, ensuring the stability of the oscillation and the accuracy of the imposed stress conditions. For each HT condition, a batch of at least 15 specimens was tested up to

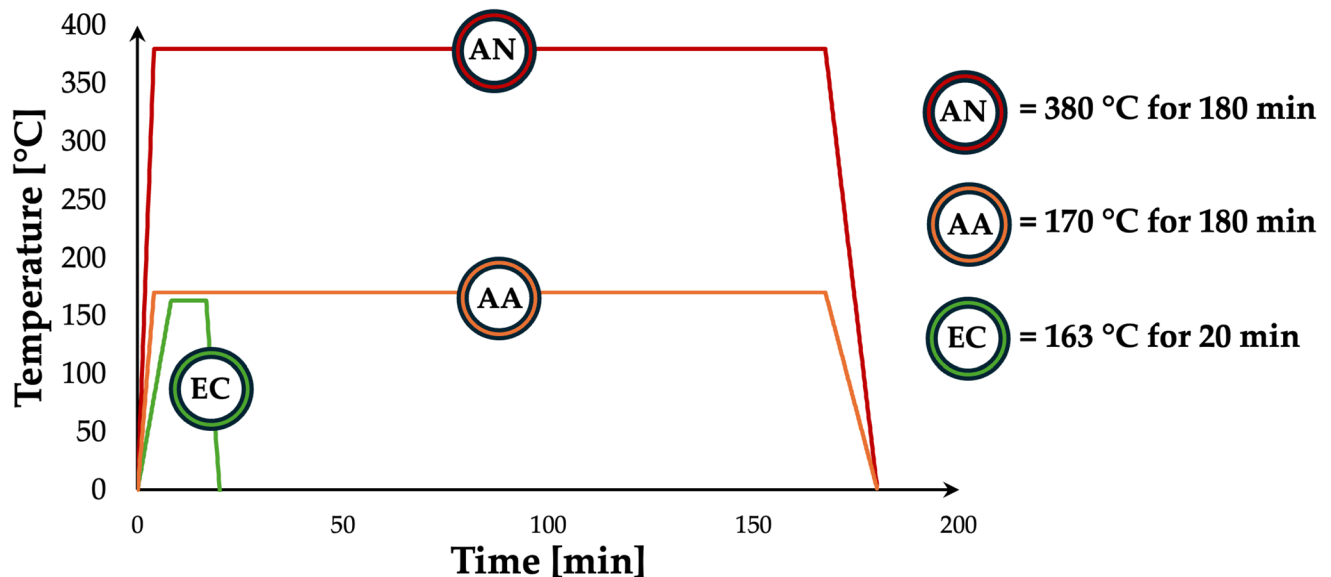


Fig. 1—Scheme of performed heat treatments: E-coating simulation (EC), artificial aging (AA), and annealing (AN).

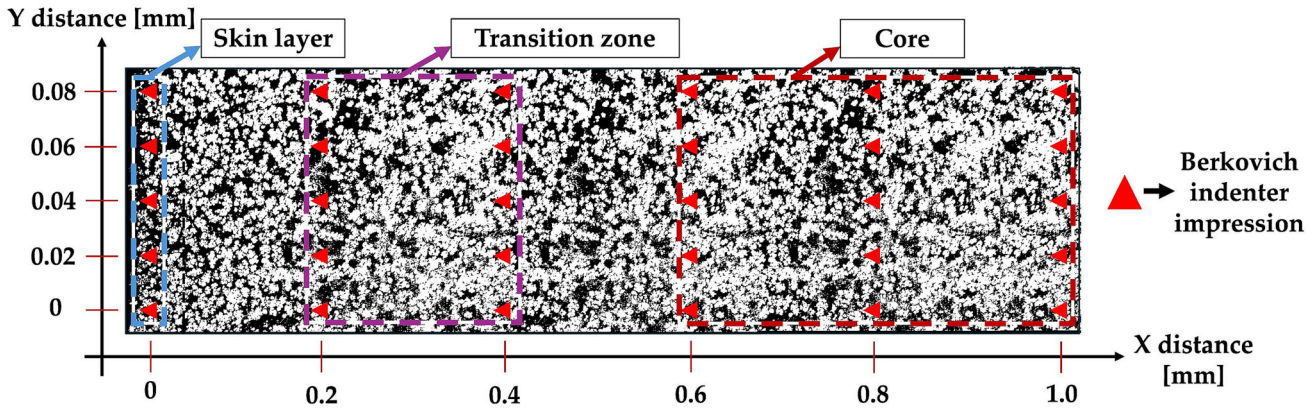


Fig. 2—Indentation matrix along the cross-section of the aluminum alloy sample. Red marks highlight Berkovich indents arranged from the skin layer (1st left column) through the transition zone (2nd to 3rd columns) to the core zone (last remaining columns).

10^7 cycles to establish a staircase curve according to ISO 12107:2012.^[36]

E. Fractographic Characterization

Fractographic analysis was performed on the fatigue fracture surfaces using SEM-FEG, to obtain high-resolution images and gain detailed insights into failure mechanisms and crack initiation sites. Secondary electron (SE) imaging was used to examine the surface morphology, while backscattered electron (BSE) imaging provided contrast based on compositional differences within the microstructure. EDS was also conducted to identify and map the elemental composition of selected areas, offering further information on phase distribution and the chemical nature of inter-metallic compounds.

F. Environmental Impact Assessment

To evaluate the environmental footprint of a potential automotive component manufactured *via* HPDC, the 2025 R1 version of Granta EduPack's Eco Audit module was employed. Key inputs such as the embodied energy and carbon emissions associated with the HR AlSi10MnMg alloy were sourced from existing literature^[37] and the environmental product declaration (EPD) of Raffmetal S.p.A.^[38] The energy consumption of the AN treatment phases was directly measured using an electric energy current meter connected to a muffle furnace (Nabertherm GmbH, Nabertherm LT 9/14, Lilienthal, Germany). Based on these measurements, energy consumption under EC and AA conditions was estimated. All data, including embodied energy [MJ/kg], climate change impact (CO₂ equivalent) [kg/kg], and the remaining parameters, are presented in the Table S1. The latter were retrieved from the internal database of Granta Selector integrated into the software.

The fatigue life of the automotive component was estimated using the Wohler curve-based fatigue life equation (Eq. [1]):

$$N = N_0 \left(\frac{\sigma_D}{\sigma_{app}} \right)^k \quad [1]$$

where k is equal to 10,^[39,40] N_0 corresponds to 10^7 cycles (reference number of cycles), σ_{app} represents the applied stress, and σ_D is the endurance limit. The calculation was performed based on the assumptions outlined in Table S1, using a constant stress amplitude of 85 MPa, taken from the material datasheet.^[41] This value corresponds to the typical fatigue resistance of the alloy and was intentionally adopted to evaluate the component under conservative, service-representative conditions. Based on these conditions, Eq. [2] was used to calculate the estimated fatigue life in terms of vehicle years of service.

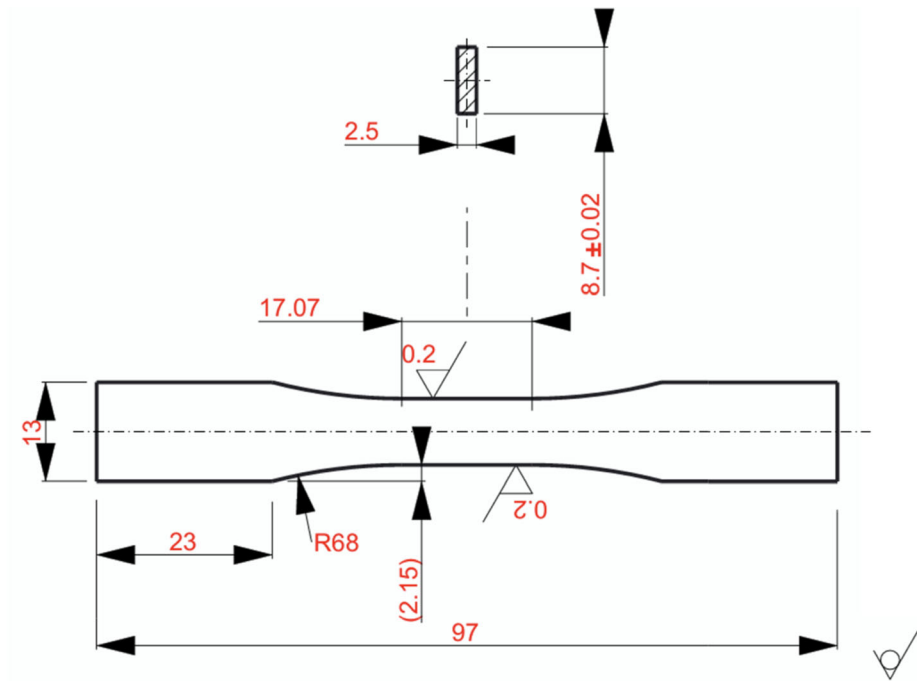
$$N_{HT} = \frac{N}{\text{distance} \cdot \text{usage} \cdot \text{cycle}} \quad [2]$$

III. RESULTS AND DISCUSSION

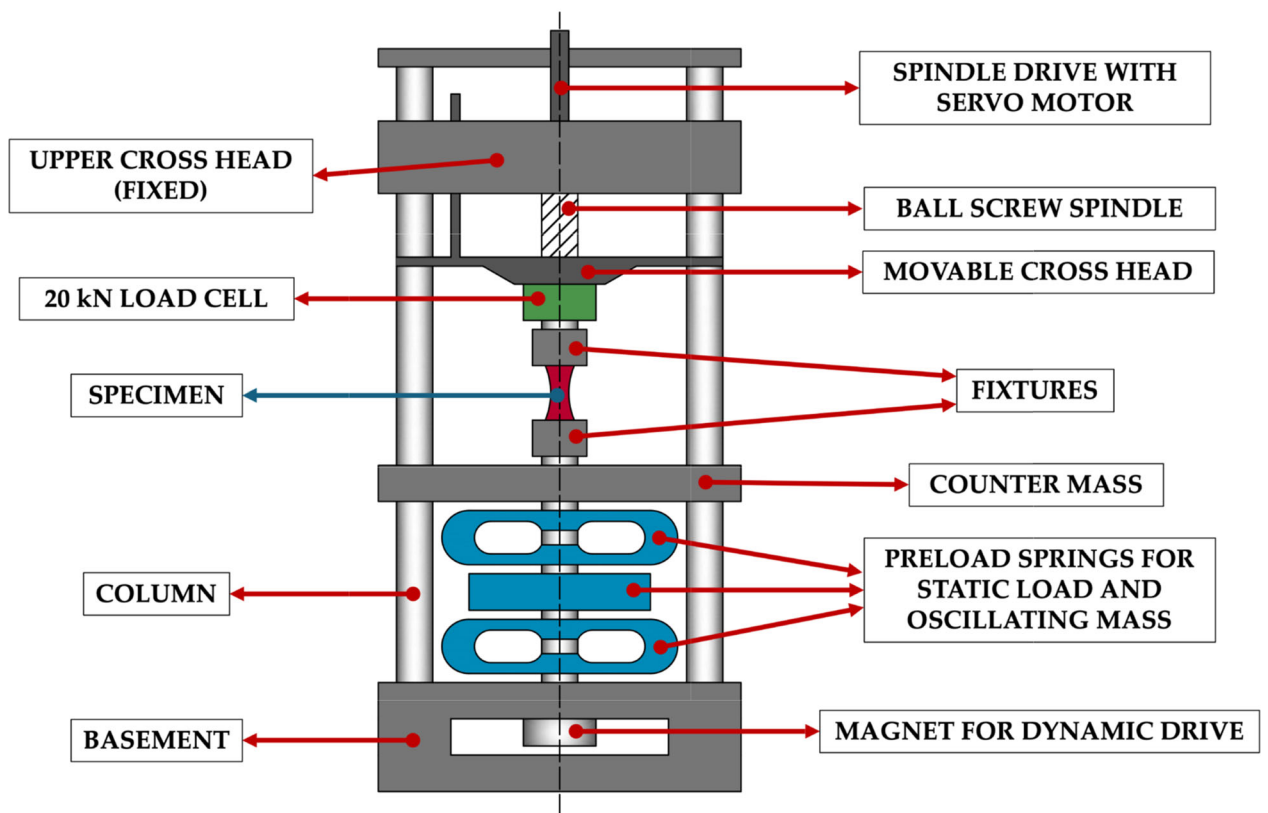
A. Microstructural Characterization

Since the EC condition involves a short thermal exposure at moderate temperature, it does not substantially alter the AP microstructure, as also reported in previous studies.^[28] Nevertheless, the EC condition was included in this work because it represents an industrially relevant post-processing step for automotive HPDC components. For this reason, EC was adopted both as a representative reference of the as-cast microstructure and as a benchmark to compare the effects of dedicated heat treatments on the fatigue response of the HR AlSi10MnMg alloy.

Figure 4(a) illustrates the regions selected for microstructural observation on the cross-section of the samples. Due to the intrinsic characteristics of the HPDC process, the formation of the skin layer is not always uniform across the casting. As highlighted by Dalai *et al.*,^[24] complex melt flow patterns and local variations in cooling rate during die filling may lead to



(a)



(b)

Figure 3—Schematic representation of (a) fatigue specimen geometry; (b) resonant testing machine, highlighting the main components (electromagnetic drive, resonance springs, preload system, and specimen clamping section).

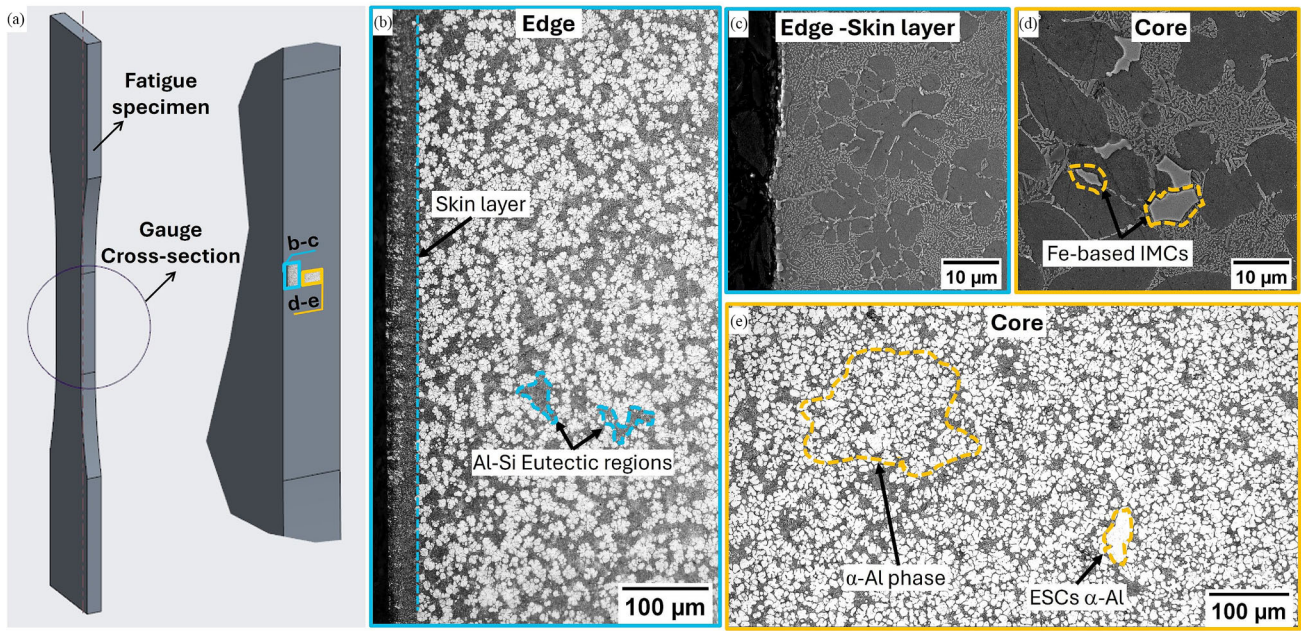


Fig. 4—(a) Schematic cross-section of a fatigue specimen in HR AlSi10MnMg alloy with a surface skin layer, indicating the regions selected for microstructural characterization: (b, c) near-surface region, showing a refined skin layer with a high fraction of eutectic Al-Si; (d, e) core region, characterized by a higher fraction of primary α -Al phase.

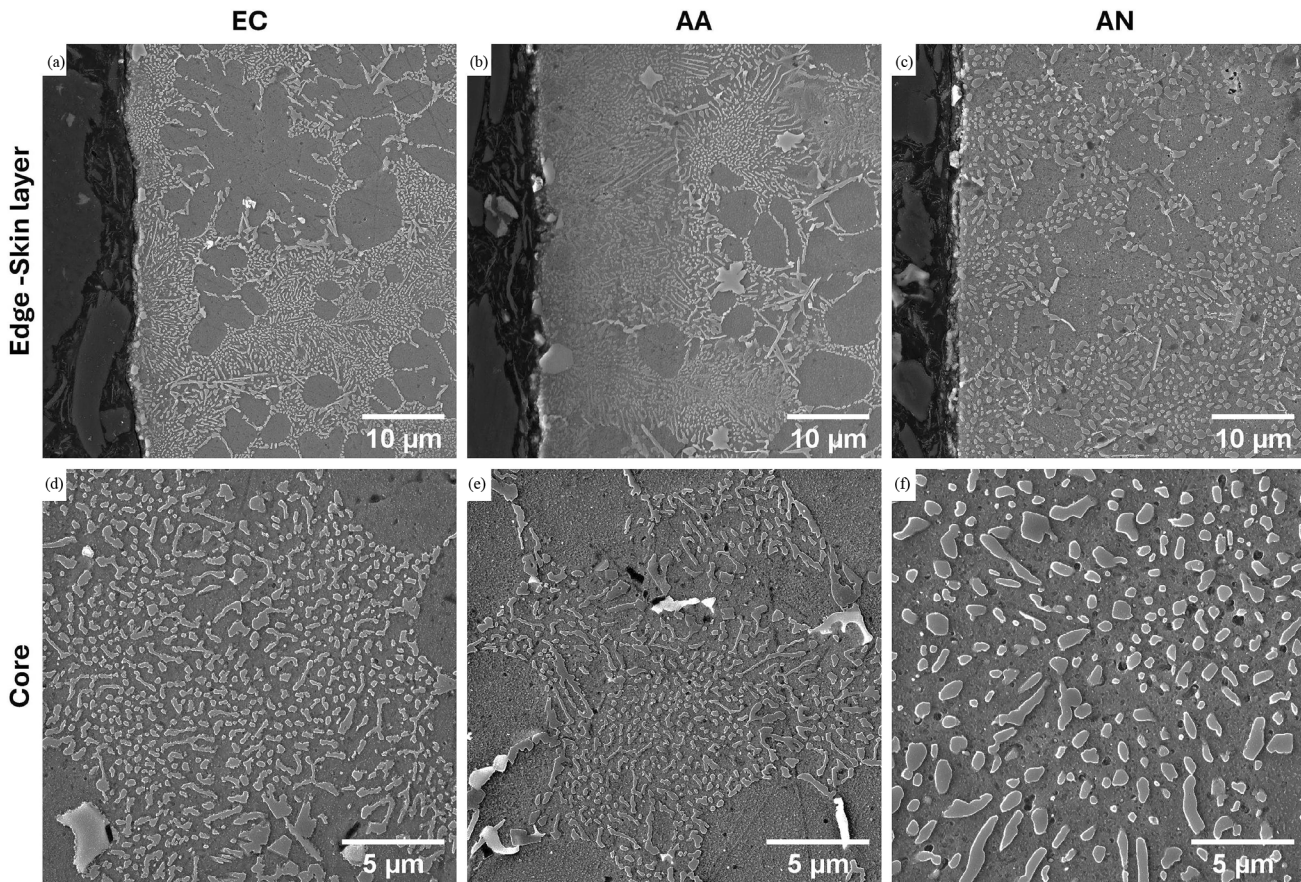


Fig. 5—SEM images focusing on the skin layer (a) through (c) and the core region (d) through (f) for an AlSi10MnMg alloy under the following heat treatment conditions: (a, d) EC; (b, e) AA; (c, f) AN.

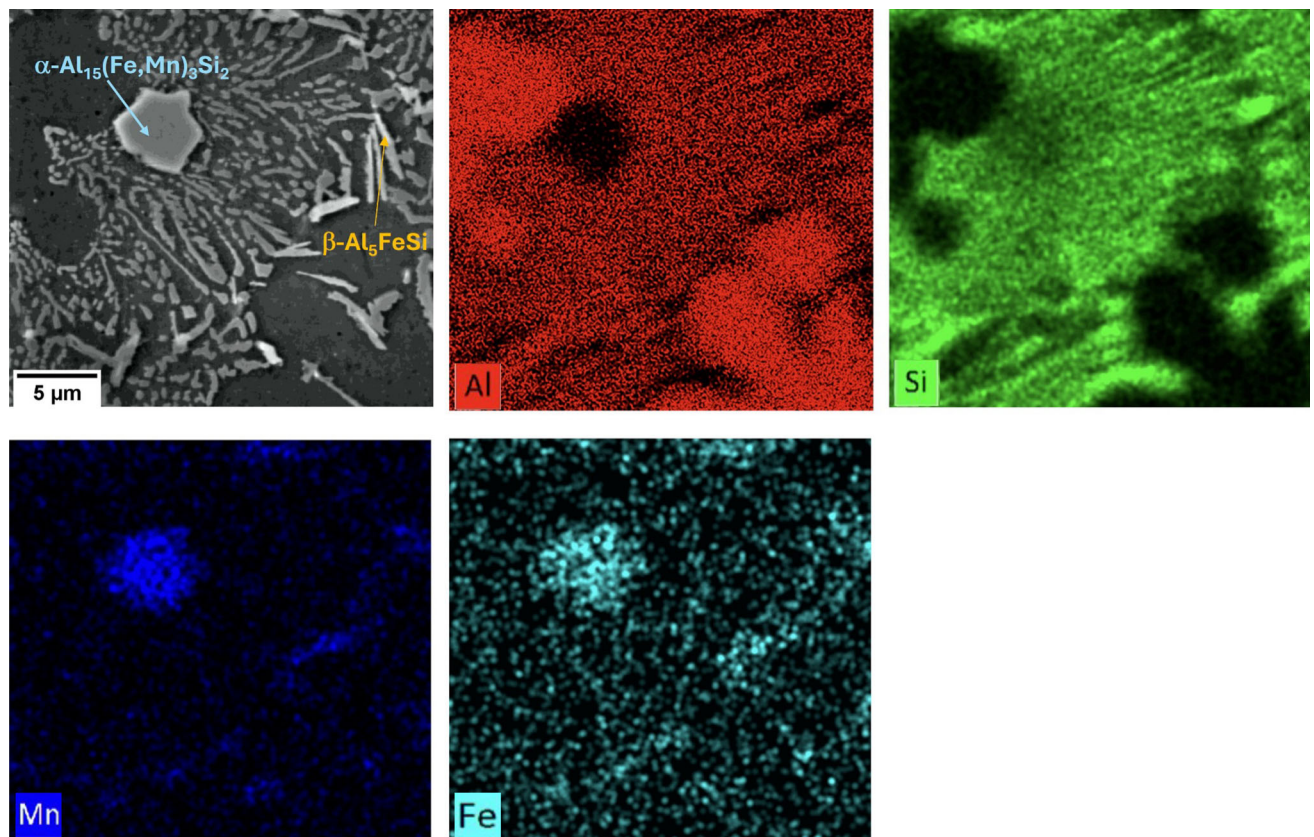


Fig. 6—SEM image and EDS elemental mapping of α - and β -Fe-based intermetallic compounds in the microstructure of the HR AlSi10MnMg alloy in the EC condition.

the formation of a discontinuous or locally developed skin layer. Nevertheless, when the skin was present, the cross-section exhibited a heterogeneous and layered morphology in which two distinct zones can be identified: the skin layer, located near the surface [Figures 4(b) and (c)], and the core region [Figures 4(d) through (e)]. The skin layer was characterized by a highly refined microstructure with a high fraction of eutectic Al–Si, while the core region showed a higher fraction of primary α -Al phase. The skin layer represents the first region to solidify, as it formed in direct contact with the die wall. This layer extended to a depth of less than 100 μm , corresponding to approximately 5 pct of the specimen thickness, and was predominantly characterized by eutectic Al–Si and highly refined α -Al grains. Such a fine microstructure resulted from the high cooling rates achieved during HPDC (generally in the range 50 to 125 $^{\circ}\text{C/s}$ [42,43]), which promoted rapid nucleation and limited growth.

Moving toward the center, the core region exhibited a coarser and more heterogeneous microstructure, consisting of a mixture of α -Al dendritic or rosette-like grains surrounded by the eutectic Al–Si, together with larger externally solidified crystals (ESCs) formed inside the shot sleeve, as described by Jiao *et al.* [22] This inner region thus presented a higher fraction of primary α -Al phase and a more open interdendritic network.

Although post-processing heat treatments did not significantly alter the thickness of the skin layer, they strongly influenced the microstructural morphology across the section, including the skin layer itself. Specifically, the eutectic Si particle size was about $0.66 \pm 0.06 \mu\text{m}$ after EC [Figures 5(a) through (d)]. Due to the limited thermal exposure during the AA, the size of eutectic Si particles did not change significantly, setting at $0.58 \pm 0.02 \mu\text{m}$ [Figures 5(b) through (e)]. In contrast, after AN, it increased by approximately 25 pct [Figures 5(c) through (f)], confirming thermally induced coarsening. Moreover, eutectic Si particles exhibited a progressive spheroidization, with roundness increasing from 0.57 ± 0.03 after EC to 0.63 ± 0.03 after AN. Further details on the quantitative microstructural characterization of the bulk region of the same HPDC recycled AlSi10MnMg component are available in previous studies by the authors. [28,29]

The microstructure of the HR AlSi10MnMg alloy, in addition to the primary α -Al matrix and the eutectic Al–Si constituent, typically contains Fe-rich intermetallic compounds, mainly identified as $\alpha\text{-Al}_{15}(\text{Fe,Mn})_3\text{Si}_2$ and $\beta\text{-Al}_5\text{FeSi}$. These phases are shown in Figure 6, which reports the SEM micrograph and corresponding EDS elemental maps. The local enrichment in Fe, Mn, and Si detected in these regions supports the identification of these intermetallic phases. From a morphological standpoint, the α -phase appeared as compact polyhedral

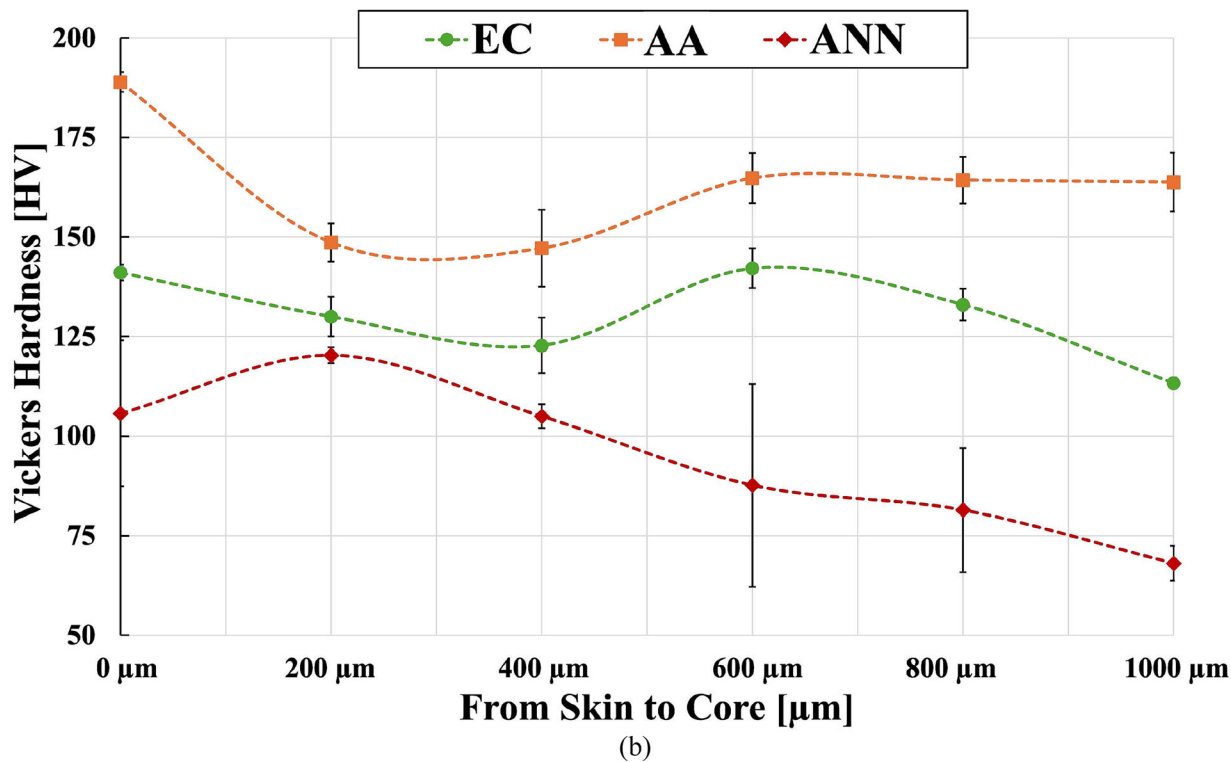
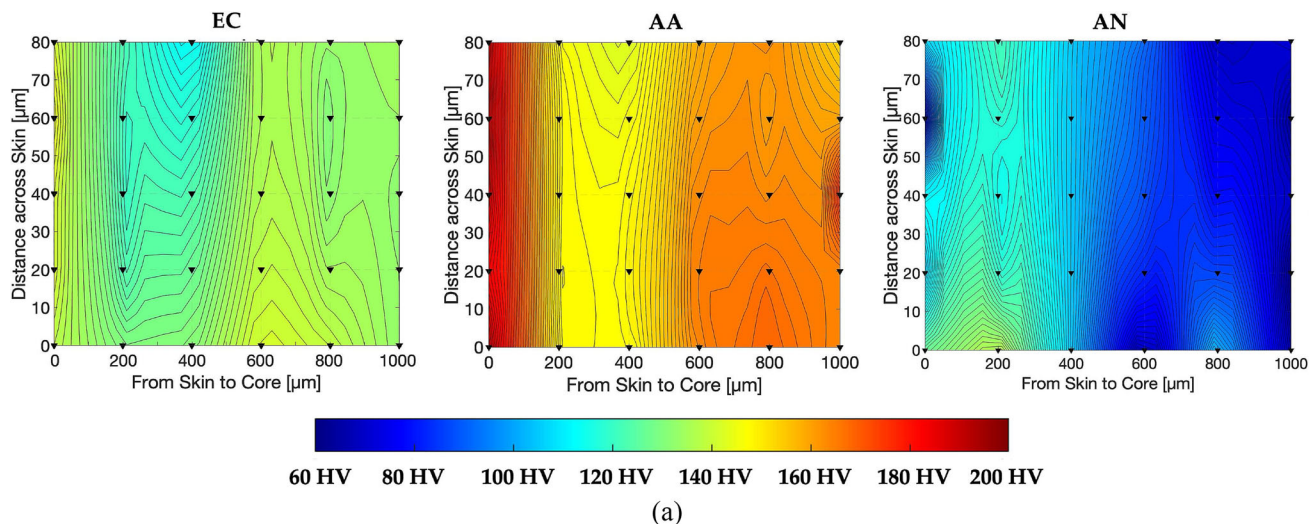


Fig. 7—Nanohardness distribution across die-cast samples after EC, AA, and AN treatment: (a) hardness maps highlighting the evolution of the skin–core hardness gradient, with triangles indicating the tested indentation points; (b) combined hardness profiles as a function of the distance from the skin layer toward the core.

particles, while the β -phase exhibited a characteristic acicular morphology.

According to Scheil–Gulliver simulations for this alloy,^[44] the presence of a minor fraction of Mg_2Si phase was also expected. During AA, this phase may partially dissolve into the Al matrix and subsequently reprecipitate as finely dispersed nanosized precipitates, as reported in Reference 45, thereby contributing to precipitation strengthening.

The coexistence of $\alpha\text{-Al}_{15}(\text{Fe},\text{Mn})_3\text{Si}_2$ and $\beta\text{-Al}_5\text{FeSi}$ phases can be attributed to the combined effect of alloy composition and the non-equilibrium solidification conditions typical of HPDC. In recycled Al–Si–Mg alloys, the formation and morphology of Fe-rich intermetallics are strongly influenced by local cooling rate, solute segregation, and the presence of Fe-correcting elements. In this regard, Mn promotes the formation of compact $\text{Al}(\text{Fe},\text{Mn})\text{Si}$ phases, reducing the tendency to form acicular $\beta\text{-Al}_5\text{FeSi}$; nevertheless, local variations in

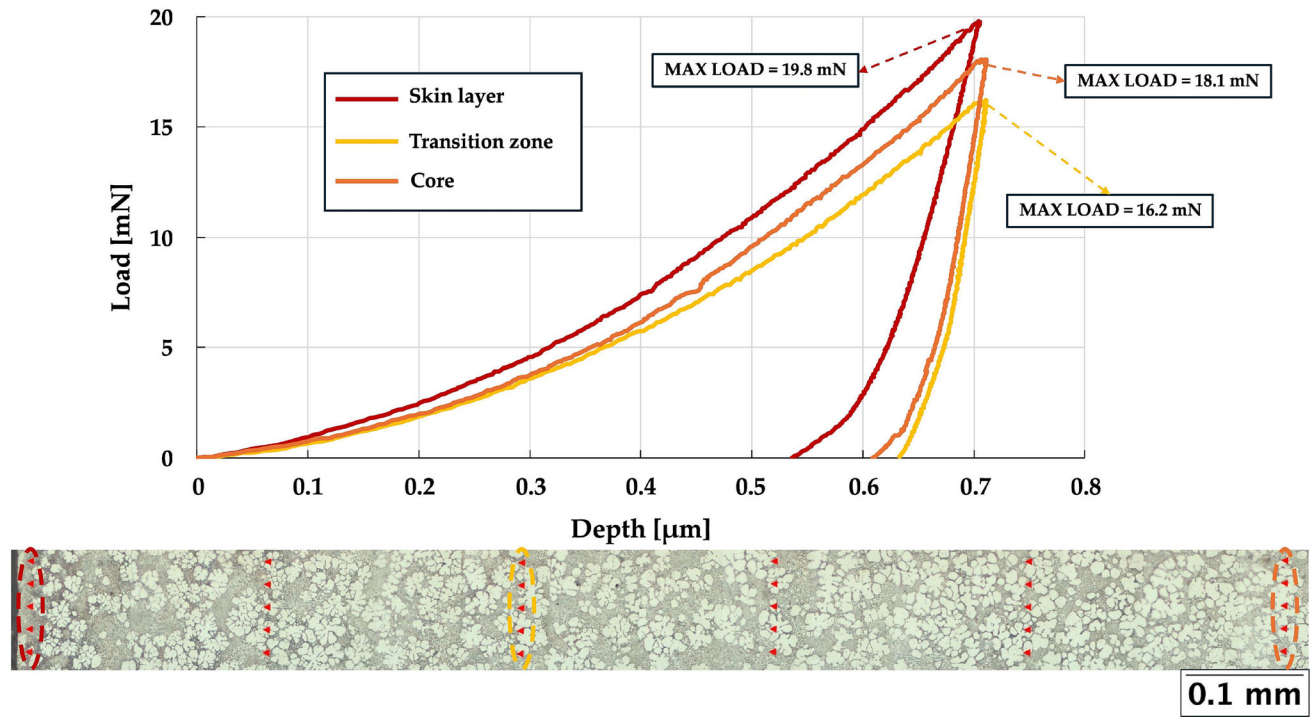


Fig. 8—Load–depth curves obtained by nanoindentation across the skin, transition zone, and core of the HR AlSi10MnMg alloy in the AA condition. The skin layer exhibits the highest resistance to indentation, followed by the core and the transition zone, confirming the presence of a mechanical gradient consistent with the hardness distribution.

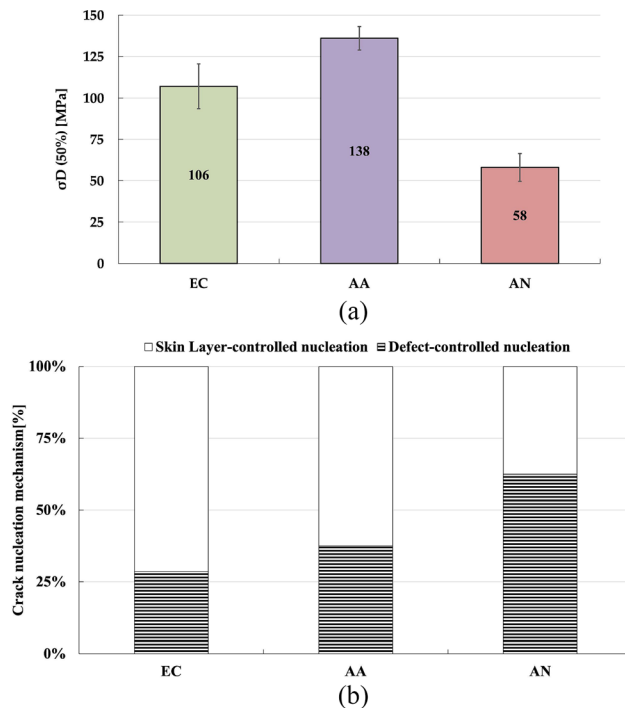


Fig. 9—(a) σ_D (50 pct) values of the HR AlSi10MnMg alloy under different heat treatment conditions: E-coated (EC), artificially aged (AA), and annealed (AN); (b) distribution of the observed crack nucleation mechanisms, distinguishing between skin layer-controlled and defect-controlled nucleation.

solidification conditions may still promote the coexistence of both phases, as also reported in recent studies on Fe-bearing phase evolution in Al–Si-based alloys.^[46,47]

B. Instrumented Indentation Tests

The evolution of HIT across the HPDC specimens is shown in the nanohardness maps in Figures 7(a) and (b).

After EC, the HIT profile showed a harder skin region (141 ± 2 HV), followed by a decrease in the transition zone (117 ± 8 HV) and a partial recovery toward the core (129 ± 2 HV), reflecting the progression from eutectic-rich skin to more dendritic regions. Following AA, the skin hardness increased to 189 ± 2 HV (+ 34 pct compared to EC), while a clear gradient was observed across the section, decreasing to 148 ± 7 HV in the transition zone and partially recovering to 164 ± 6 HV in the core. Similar surface–core hardness gradients have been reported in Reference 48, where the surface layer exhibits a stronger age-hardening compared to the core region. This trend is further confirmed by the average loading–unloading curve comparison reported in Figure 8, showcasing the same mechanical hierarchy observed in the HIT maps: the skin layer exhibited the highest resistance to indentation, followed by the core and the transition zone.

After AN, hardening phases coarsen, leading to a softening across the whole section, with skin hardness lowering to 106 ± 18 HV (– 44 pct compared to AA) and core lowering to 88 ± 10 HV. The hardness gradient

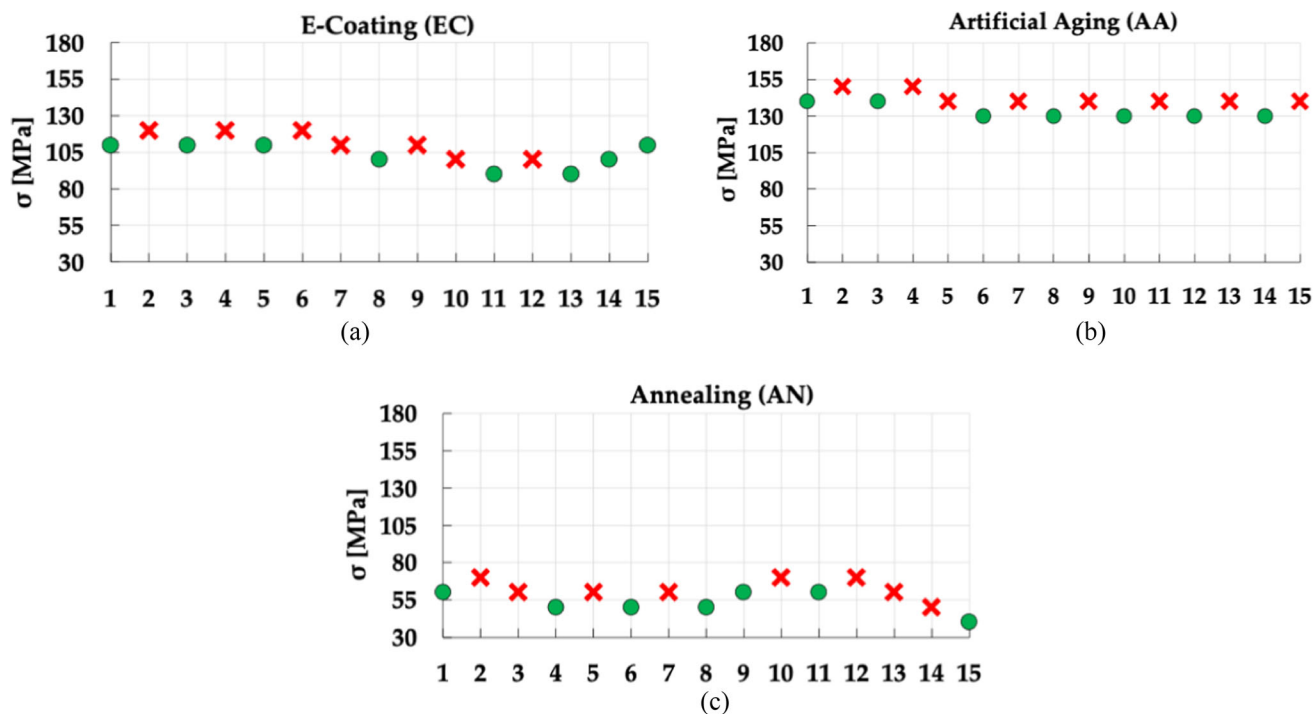


Fig. 10—Staircase fatigue test results for the three investigated conditions: (a) EC, (b) AA, and (c) AN.

disappears: the skin layer loses its peculiar mechanical advantage, and the alloy shows a uniformly lower hardness.

C. Fatigue Tests

The results of the fatigue assessment are summarized in Figure 9, which reports the σ_D (50 pct) limits for each HT condition. Figure 10 complements this information by displaying the staircase curves for the three investigated conditions, including stress amplitudes and failure/run-out patterns.

EC exhibited not only an intermediate fatigue strength (106 ± 12 MPa), reflecting the beneficial contribution of the refined skin layer, but also the negative effect of a softer core. In addition, EC showed a larger data scatter compared with the other HTs. This behavior can be attributed to the short thermal exposure associated with the treatment, which did not promote a homogeneous precipitation of strengthening phases throughout the sampled material. In addition to local differences in precipitation response, the local variability of the as-cast skin layer may have contributed to the scatter observed in the staircase fatigue results, as also reflected by the distribution of crack nucleation mechanisms reported in Figure 9(b). Therefore, local variations in microstructure and mechanical properties may develop between the skin layer and the core, leading to a more heterogeneous fatigue response.

AA led to the highest fatigue strength (138 ± 4 MPa), owing to the precipitation of strengthening phases across the section. This precipitation hardening

increases the overall resistance to cyclic plastic deformation, resulting in the best fatigue performance among the investigated conditions.

Conversely, after AN, the fatigue strength dropped sharply to 58 ± 8 MPa. As shown in Figures 5(c) through (f), at the selected annealing temperature the eutectic Si particles undergo coarsening, leading to a progressive homogenization of the microstructure and a reduction of its strengthening capability. In addition, the annealed condition is not supported by strengthening precipitates, which further contributes to the observed decrease in fatigue resistance. This microstructural evolution is consistent with the nanoindentation results (Figure 8), which reveal a general decrease in hardness in both the skin layer and the core region, resulting in hardness values comparable to those measured in the transition zone and indicating the disappearance of the skin–core hardness gradient.

D. Fractographic Characterization

While nanoindentation maps (Figure 7) and fatigue results [Figure 9(a)] indicate that heat treatment strongly influences the fatigue behavior of HR AlSi10MnMg HPDC castings through the hardness distribution across the section, fractographic analysis highlights the key role of the surface skin layer in controlling fatigue crack nucleation. When the skin layer is present, the hardness gradient between the refined surface region and the underlying bulk tends to shift crack nucleation toward the skin–transition zone interface. In this case, the skin layer behaves as a protective surface layer that shifts crack nucleation from the external surface toward the skin–transition zone interface. A similar interaction

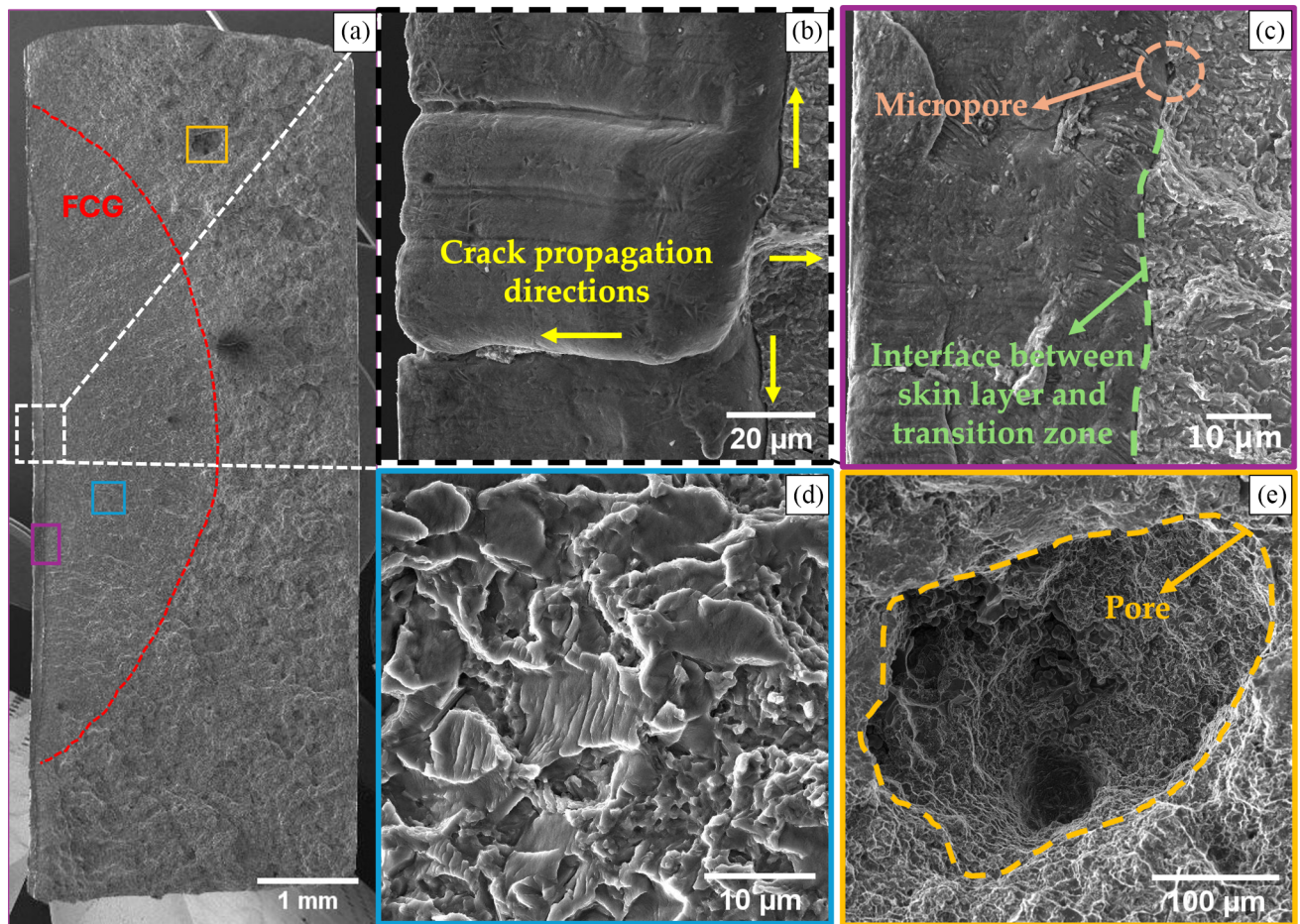


Fig. 11—(a) Low magnification SEM image of the fracture surface of a fatigue specimen under AA heat treatment condition (7.4×10^4 cycles at 140 MPa) and high magnification images of: (b, c) crack initiation sites, (d) crack propagation region, i.e., fatigue crack growth (FCG); and (e) overload fracture region.

between the skin layer and the underlying microstructure was previously reported under static tensile loading by Jiao *et al.*,^[8] who observed that the stronger skin layer initially resists deformation while the bulk yields earlier. The resulting mismatch in mechanical response promotes microcrack formation at the interface, which can subsequently act as a crack initiation site, as shown in Figure 11.

In this case, crack nucleation occurs within the interfacial region between the skin layer and the transition zone [Figures 11(b) and (c)], where local stress concentrations arise from the hardness gradient and the presence of microstructural heterogeneities, such as interfacial discontinuities and micropores. This suggests a distributed nucleation process along the skin–transition zone interface. The fatigue crack subsequently penetrates the skin layer and propagates through the bulk material until final fracture occurs by overload [Figures 11(d) and (e)].

When the skin layer is absent, fatigue cracks nucleate directly at the external surface, typically from casting defects, as shown in Figure 12.

As shown in Figures 12(b) and (c), crack nucleation occurred from a surface-connected shrinkage cavity

intercepted by the specimen side during sample extraction. After nucleation, the crack propagated through the bulk material, forming a fatigue crack growth (FCG) region characterized by fatigue striations [Figure 12(d)]. Failure finally occurred by overload, as indicated by the presence of dimples on the fracture surface [Figure 12(e)].

The two crack initiation conditions identified from the fractographic analysis are summarized in the schematic representation reported in Figure 13, and the relative occurrence of these mechanisms for the analyzed fracture surfaces is reported in Figure 9(b).

Overall, the reported model indicates that fatigue crack nucleation is strongly affected by local surface and near-surface microstructural conditions. In the presence of an effective skin layer, crack nucleation tends to occur at the skin–transition zone interface, whereas, when the skin layer is absent or locally discontinuous, casting defects can become preferential crack nucleation regions. This behavior reflects the locally developed nature of the skin layer in industrial HPDC components, whose thickness and continuity depend on metal flow, die wall contact, and local heat extraction during solidification. After nucleation, crack propagation

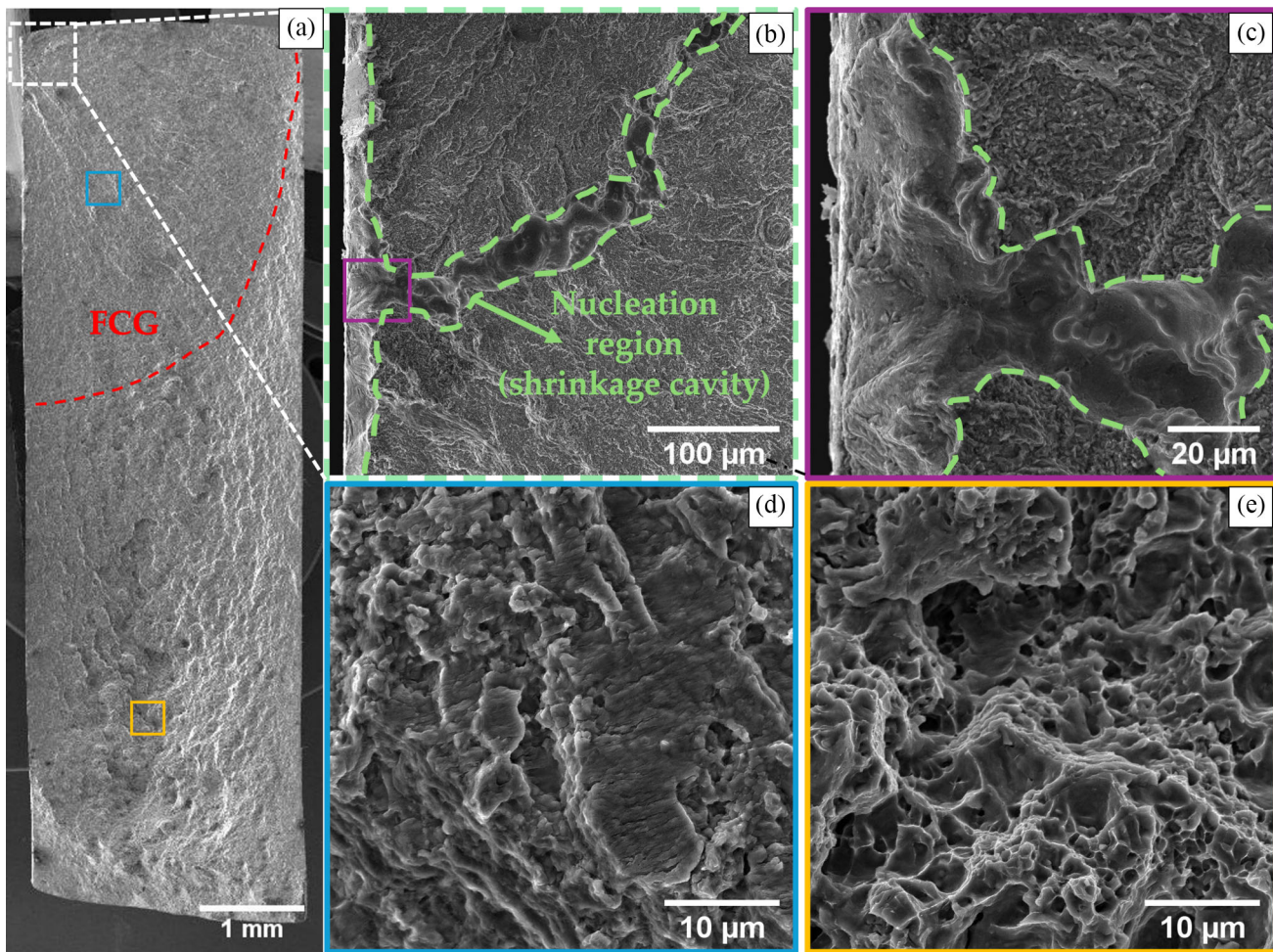


Fig. 12—(a) Low magnification SEM image of the fracture surface of a fatigue specimen under AN heat treatment condition (2.2×10^5 cycles at 60 MPa) and high magnification images of: (b) crack nucleation region, (c) detail of the interconnection of the shrinkage cavity between the fracture surface and the specimen side; (d) crack propagation region, *i.e.*, fatigue crack growth (FCG); and (e) overload fracture region.

proceeds through the bulk material, where interactions with microstructural heterogeneities, such as Fe-based intermetallic compounds (which were detected along the fracture path by EDS maps shown in Figure S1) locally influence the crack path. Overall, these observations highlight that the fatigue response is governed by the combined action of local surface conditions, which control crack nucleation through the effectiveness and continuity of the skin layer, and the heat treatment-dependent hardness of the bulk material, which influences crack propagation.

E. Environmental Impact Assessment

The energy required for each stage of the component life cycle, including raw material production, manufacturing (with HPDC and subsequent HTs), transportation, and in-service use, provides a comprehensive picture of the contributions to total energy consumption and associated CO₂ emissions. The results, shown in Figure 14, highlight the environmental impact differences between components made from the HR AlSi10MnMg alloy under two different heat treatment

conditions: EC and AA. The AN condition was excluded from the environmental impact assessment, as this condition is usually applied for components operating under moderate cyclic loading where good electrical and thermal conductivity is required.^[28]

Based on the results reported in Figure 9(a) and the assumptions detailed in Table S1, it can be concluded that a potential automotive component manufactured by HPDC in HR AlSi10MnMg alloy, with a mass of 1 kg, and subjected to cyclic loading of 85 MPa, represents a severe fatigue condition. Under these loading conditions, the estimated fatigue life, measured through Eq. [2], strongly depends on the HT. A component in the EC condition is predicted to endure for approximately 8.5 years, whereas the same component subjected to AA condition would reach a fatigue life of around 91.5 years. Given that, the average operational life of a vehicle is approximately 15 years, the EC-treated component, subject to that cyclic loading, would not survive the full-service duration and would therefore require at least one replacement. This replacement scenario effectively doubles the environmental impact, as illustrated by the hatched sections in

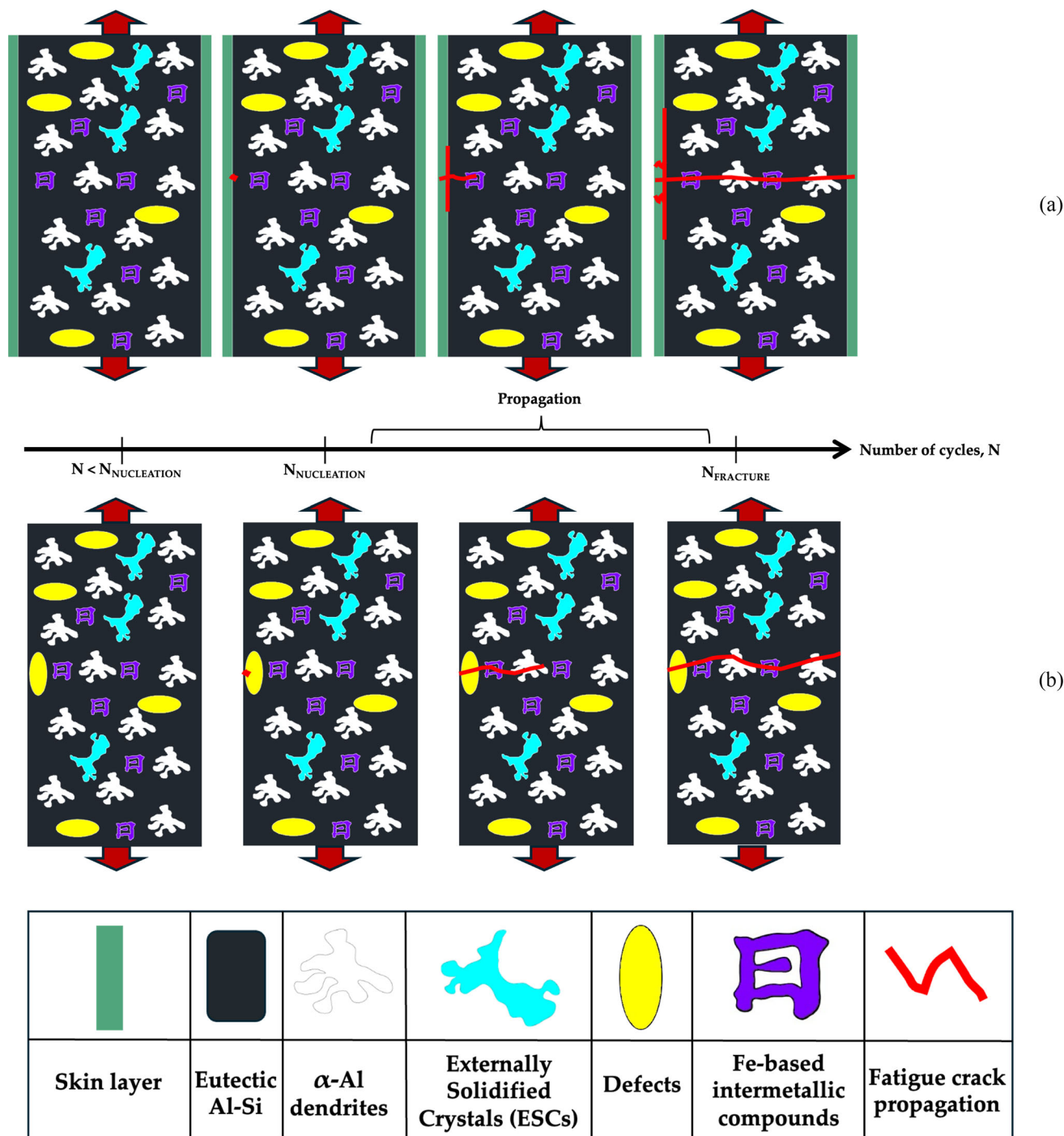


Fig. 13—Schematic representation of fatigue crack nucleation and propagation mechanisms in HR AlSi10MnMg HPDC alloy as a function of the presence of the surface skin layer: (a) subsurface crack nucleation at the skin–transition zone interface when the skin layer is mechanically effective; (b) surface crack nucleation from casting defects when the skin layer is absent.

Figure 14, which highlight duplicated contributions. From an environmental standpoint, when this additional replacement is accounted for, the AA-treated component results in a 17.44 pct reduction in embodied energy and a 23.33 pct reduction in climate change potential (expressed as kg CO₂ equivalent per kg of component), compared to the EC condition. Although the AA treatment, when considered in isolation, entails

a higher energy consumption due to the aging step (Table S1), it leads to a significant improvement in fatigue performance, increasing the fatigue strength σ_D by approximately 30 MPa. This enhancement in mechanical durability justifies the initial energy investment, as it ensures a service life well beyond that of the vehicle, thereby eliminating the need for replacement and reducing long-term environmental impact.

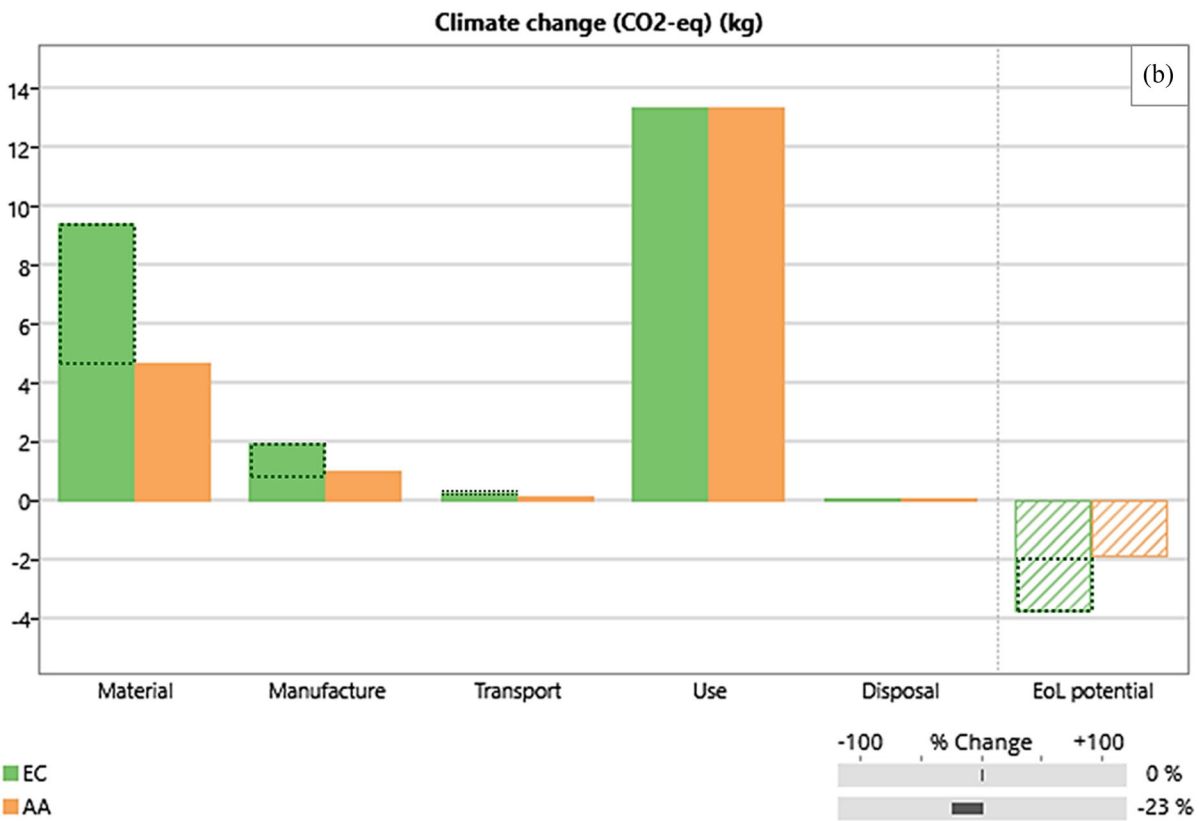
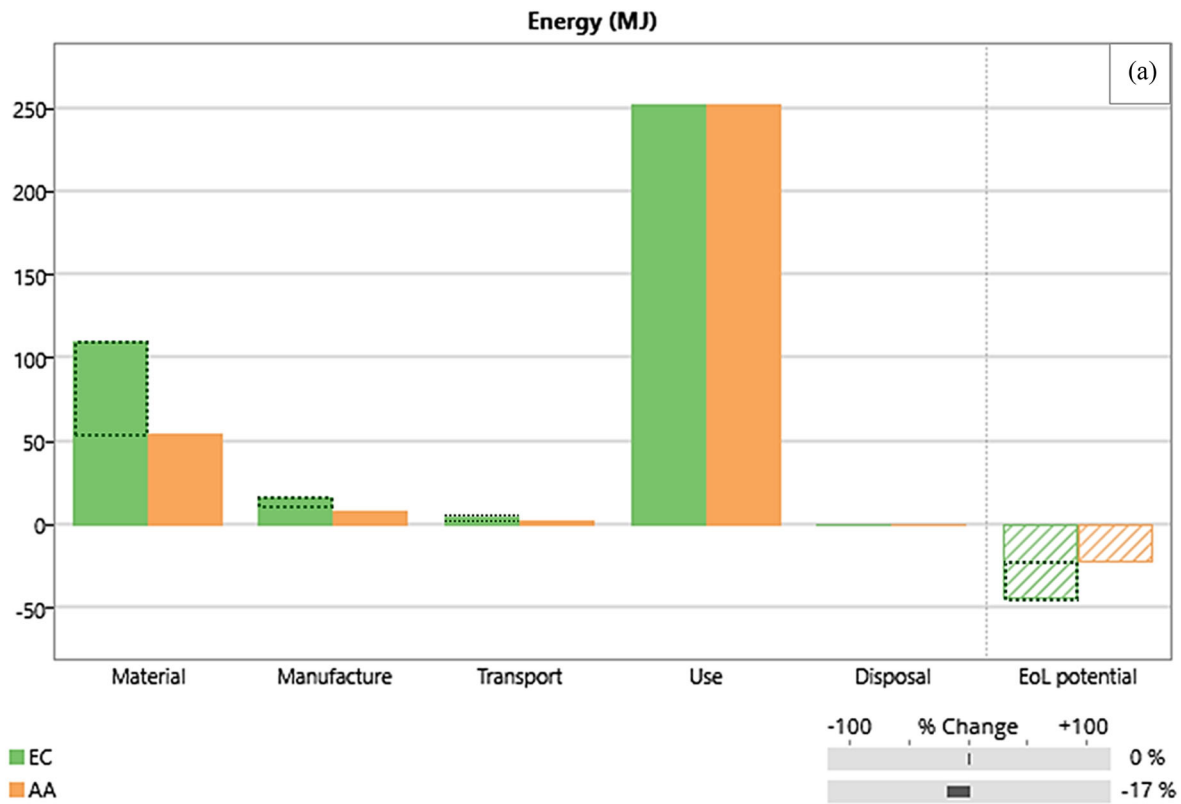


Fig. 14—Eco Audit of an automotive component produced through AlSi10MnMg HPDC alloy: (a) energy; (b) CO₂ footprint.

For both conditions, the analysis clearly shows that the most significant contribution to environmental burden arises from the production of the raw material, which far exceeds the impact of the manufacturing phase itself (including HPDC and HTs). The transportation stage has a negligible influence, assuming distribution distance of about 1200 km from the production site (Table S1).

IV. CONCLUSIONS

The influence of different heat treatments such as E-coating simulation (EC), artificial aging (AA), and annealing (AN) on a 75 pct recycled (HR) AlSi10MnMg HPDC automotive component was systematically investigated in terms of microstructure, fatigue behavior, failure mechanisms, and environmental impact. The following observations can be highlighted:

- The HR AlSi10MnMg alloy exhibited a heterogeneous cross-sectional microstructure characterized by a locally developed eutectic-rich skin layer and a coarser dendritic core. Due to the intrinsic characteristics of the HPDC process, the skin layer was not always uniformly developed along the surface.
- Nanoindentation maps revealed a pronounced hardness gradient across the section under EC and AA conditions, whereas the AN treatment led to a microstructural coarsening and a more homogeneous hardness distribution.
- Fatigue tests showed that the AA condition provides the highest fatigue strength (138 ± 4 MPa), followed by EC (106 ± 12 MPa) and AN (58 ± 8 MPa), highlighting the strong influence of heat treatment on the overall fatigue performance.
- Fractographic observations indicate that fatigue crack nucleation was strongly influenced by the local effectiveness and continuity of the skin layer. When present, the skin layer shifted crack nucleation toward the skin-transition zone interface, while its absence promoted nucleation from casting defects. Crack propagation, instead, was mainly governed by the hardness distribution induced by heat treatment within the bulk material.
- The environmental impact assessment showed that, despite a higher processing energy consumption, AA can extend component lifetime and reduce the frequency of replacements compared to EC, thereby lowering the overall life-cycle environmental impact up to about 20 pct reduction in embodied energy and CO₂ emissions.

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AUTHOR CONTRIBUTIONS

Cristian Cascioli, Riccardo Arcaleni, Elena Mingotti, Luca Girelli, Luca Lorenzetti: methodology, investigation, writing—original draft preparation, writing—review and editing. Alessandro Morri, Lorella Ceschini, Annalisa Pola: conceptualization, validation, supervision. All authors have read and agreed to the published version of the manuscript.

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CONFLICT OF INTEREST

The authors declare that they do not have any financial or personal relationships with other people or organizations that could have inappropriately influenced this study.

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