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Microstructure and Mechanical Properties of Al-Ce and Al-Ce-Cr Cast Alloys: Effects of Cr and Solidification Conditions

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Abstract

Lightweight Al alloys capable of retaining mechanical performance at elevated temperatures are increasingly relevant for transportation and structural applications. Al-Ce alloys are particularly attractive because of their good castability and thermally stable Ce-containing intermetallics. However, how Cr addition modifies phase formation and mechanical behavior under rapid solidification remains insufficiently understood. This study investigates the processing–microstructure–property relationships of Al–13Ce and Al–10Ce–1Cr (wt%) alloys produced by centrifugal solidification (CS), with directional solidification (DS) used for comparison. CS was performed using a stepped Cu mold with bulk diameters of 5–11 mm, and the resulting alloys were characterized by SEM/EDS, XRD, Vickers microhardness, and compression testing. CS produced markedly refined microstructures, with dendritic spacings of approximately 4.0–5.4 μm . The Al–13Ce alloy exhibited a cellular-to-dendritic transition corresponding to samples having smaller diameters, whereas Al–10Ce–1Cr retained a dendritic morphology throughout all conditions. The Cr-containing alloy generally exhibited higher compressive and yield strengths, reaching 282 and 143 MPa, respectively, compared with 253 and 119 MPa for Al–13Ce in the same 11-mm CS condition. The Hollomon analysis revealed a higher strength coefficient (K) but lower strain-hardening exponent (n) for Al–10Ce–1Cr, indicating higher resistance to plastic flow but a reduced strain-hardening capacity, consistent with the presence of primary $\text{Al}_{20}\text{CeCr}_2$ particles. Overall, the compressive response was governed not only by microstructural refinement but also by phase constitution, intermetallic distribution, and particle/matrix interactions, highlighting the combined roles of alloy chemistry and solidification route in tailoring Al-Ce-based alloys.

Keywords: Al–Ce alloys; Centrifugal solidification; Rapid solidification; Intermetallic compounds; Compressive behavior.

1 Introduction

Lightweight alloys that retain their mechanical performance at high temperatures are highly desirable for transportation and structural applications [1,2]. In this context, Al-based alloys containing rare-earth elements (REEs) are promising candidates owing to their low density, suitable mechanical properties, and thermal stability [3,4]. Among them, Al-Ce alloys are of particular interest because of their good castability, low coarsening rate of their second-phase particles, and retention of mechanical properties at high temperatures [5,6]. The relatively high abundance of Ce and its presence in REE-bearing ores, such as bastnäsite and monazite, further support the development of these alloys [7]. In terms of industrial processing, Al-Ce alloys have been used to produce automotive components such as air-cooled engine cylinder heads, rotary engine rotors, and high-performance pistons [5,8]. This demonstrated potential for large-scale adoption in automotive manufacturing reinforces their practical importance. Under a widespread-adoption scenario for Al-Ce alloys in high-efficiency spark-ignition engine applications, Ce demand has been projected to reach up to approximately 23,200 t per year in the United States by 2045 [9]. Therefore, establishing processing-microstructure-property relationships in Ce-containing Al alloys is fundamental for optimizing their performance to meet current and prospective technological demands.

From a microstructural perspective, the strengthening response of Al-Ce alloys is strongly influenced by the intermetallic compounds (IMCs) formed during solidification, particularly $\text{Al}_{11}\text{Ce}_3$, which is characterized by limited coarsening even after prolonged exposure to 500 °C [10,11]. This microstructural stability, combined with the compositional flexibility of Al-Ce alloys, allows their properties to be tailored through alloying additions that modify the formation, morphology, and distribution of Ce-containing phases while promoting the formation of new IMCs

[6]. Ekapruta et al. [12] demonstrated this potential in hypereutectic Al-2Ce-1X (at.%) alloys (X = Mn, Cr, V, Mo, and W) produced by laser powder bed fusion (LPBF), reporting higher hardness in the as-fabricated condition compared with the binary Al-Ce alloy, which was mainly attributed to the formation of fine Al₂₀CeX₂ precipitates for X = Cr, V, Mo, and W.

Cr induces considerable microstructural changes in heat-resistant Al alloys owing to its low solubility and diffusivity in Al, thereby affecting solute redistribution, microstructural refinement, and recrystallization behavior [12–14]. In Al-Ce alloys, Cr addition has been shown to modify the solidification microstructure through the formation of primary AlCeCr IMCs, which can locally disrupt cooperative eutectic growth and promote eutectic coarsening under as-cast conditions [15]. The mechanical implications of these microstructural changes for Al-Ce(-Cr) alloys, however, require further investigation. In this regard, compressive testing can reveal how the morphology, size, and distribution of IMCs affect plastic deformation and strength. During high-temperature compression, Al-(5-20 wt.%)Ce alloys, with room-temperature yield strengths ranging from 60 to 100 MPa, exhibited a decrease in yield stress with increasing temperature, reaching approximately 50% and 25% of their respective room-temperature values at 400 and 500 °C, respectively [10]. Compressive creep tests conducted at 300 °C revealed that the Al-2Ce-1Cr (at.%) alloy exhibits a dislocation-creep threshold stress nearly three times that of the binary Al-2Ce (at.%) alloy, mainly due to the formation of coarsening-resistant Al₂₀CeCr₂ precipitates [12]. Despite these findings, the solidification conditions required to develop microstructures that enhance the compressive behavior of Al-Ce(-Cr) alloys remain to be fully established.

To address these gaps, centrifugal solidification (CS) allows higher cooling rates to be achieved through the adjustment of processing and geometric parameters, including mold diameter [16,17]. A previous study demonstrated that CS could produce ultrafine microstructures in Al-

based alloys at cooling rates of approximately 100 K/s or higher [18]. In as-solidified Al alloys, the microstructural length scale, commonly characterized by cellular or dendritic spacing, is primarily determined by solidification kinetics, which in turn influence the morphology and distribution of secondary phases, eutectics, and IMCs. Consequently, increasing the cooling rate can induce microstructural changes whose effects on mechanical properties extend beyond those associated with grain refinement alone [19]. CS therefore offers a suitable approach for investigating the evolution of as-solidified microstructures and their relationship with the compressive response of Al-Ce(-Cr) alloys under rapid solidification conditions.

Recently, using directional solidification (DS), De Paula et al. [20] demonstrated that the compressive response of slowly cooled Al-13Ce and Al-10Ce-1Cr alloys (wt.%) is strongly influenced by microstructural evolution across cooling rates ranging from 0.7 to 19.8 K/s. In the present work, this range of cooling rates is extended via CS to assess additional microstructural transitions under rapid solidification conditions. The same alloy compositions are processed by CS using a Cu mold with four different diameters to achieve higher cooling rates. This study aims to characterize the resulting microstructural evolution and assess its effects on microhardness and compressive response. The results obtained using CS are also compared with those previously found through DS to evaluate the influence of the solidification route. This comparison provides further insight into the processing-microstructure-property relationships of Al-Ce(-Cr) alloys and supports the development of lightweight alloys to address current and emerging industrial needs.

2 Materials and methods

The experimental procedure was structured into four successive stages, as illustrated in the flowchart shown in Figure 1. Initially, ingots of the Al-13 wt.%Ce and Al-10 wt.%Ce-1 wt.%Cr

alloys were produced in an induction furnace and cast into a graphite mold. Subsequently, the alloys were remelted and solidified via two distinct routes: upward directional solidification under transient heat extraction, and centrifugal casting in a copper mold. Samples extracted from transverse sections of both ingots were then prepared to reveal their microstructural characteristics. Finally, characterization was conducted by SEM/EDS and XRD, complemented by mechanical property evaluation through compression and hardness testing.

2.1 Preparation and production of alloy samples

The aluminum-based alloys, Al-13 wt.%Ce and Al-10 wt.%Ce-1 wt.%Cr, were produced and processed according to the methodology detailed by De Paula et al. [20], which established the melting and directional solidification conditions for these alloys. Elements with commercial purity exceeding 99.7% (Al 99.8%; Cr 99.95%) and Al-20 wt.%Ce master alloy were used, weighed on a digital balance with an accuracy of 0.01 g.

Melting was carried out in an induction furnace (Inductotherm, 50 kW, 3.2 kHz) using graphite crucibles, followed by a homogenization period of melting. After thermal stabilization, the alloys were subjected to processing under two distinct heat extraction conditions, i.e., CS and DS.

The alloys were remelted in an upward directional solidification furnace. A cylindrical AISI 310 stainless steel mold (56 mm diameter and 158 mm length) was placed over a water-cooled SAE 1020 steel base, allowing heat extraction under transient conditions. Thermal control was achieved using eight type-K thermocouples positioned along the mold, enabling cooling rates ranging from 0.7 to approximately 20 K/s. For the present study, microstructural and thermal analyses were focused on three specific cooling rates - 15.9 °C/s, 3.2 °C/s and 0.76 °C/s - selected

for being common and closely matched between the two investigated alloys, thus ensuring comparability of the results. For each of these cooling rates, the corresponding positions along the mold (with origin at the mold/water-cooled base interface) were determined: for the Al-13 wt.%Ce alloy, 5.1 mm, 20.2 mm and 70.2 mm; for the Al-10 wt.%Ce–1 wt.%Cr alloy, 9.1 mm, 24.4 mm and 69.3 mm, respectively.

Another portion of each alloy was reprocessed using a centrifugal casting furnace equipped with a copper mold in a plate configuration. The processing of the alloys by centrifugal solidification (CS) was conducted with the aim of subjecting the material to rapid solidification regimes, achieving cooling rates significantly higher than those obtained via directional solidification due to the higher heat transfer coefficients attained. The samples were melted *in situ* in an induction system under an inert argon (Ar) atmosphere, ensuring protection against oxidation of reactive elements. This process was carried out using a Linn High Therm GmbH Titancast 700-VAC system. Heating was precisely controlled until the molten metal reached a temperature of 800 °C. The copper mold contains stepped cylindrical cavities with diameters ($\varnothing$) of 5, 7, 9 and 11 mm, a configuration strategically adopted to promote different solidification rates in a single run, thereby generating distinct levels of microstructural refinement for comparative analysis.

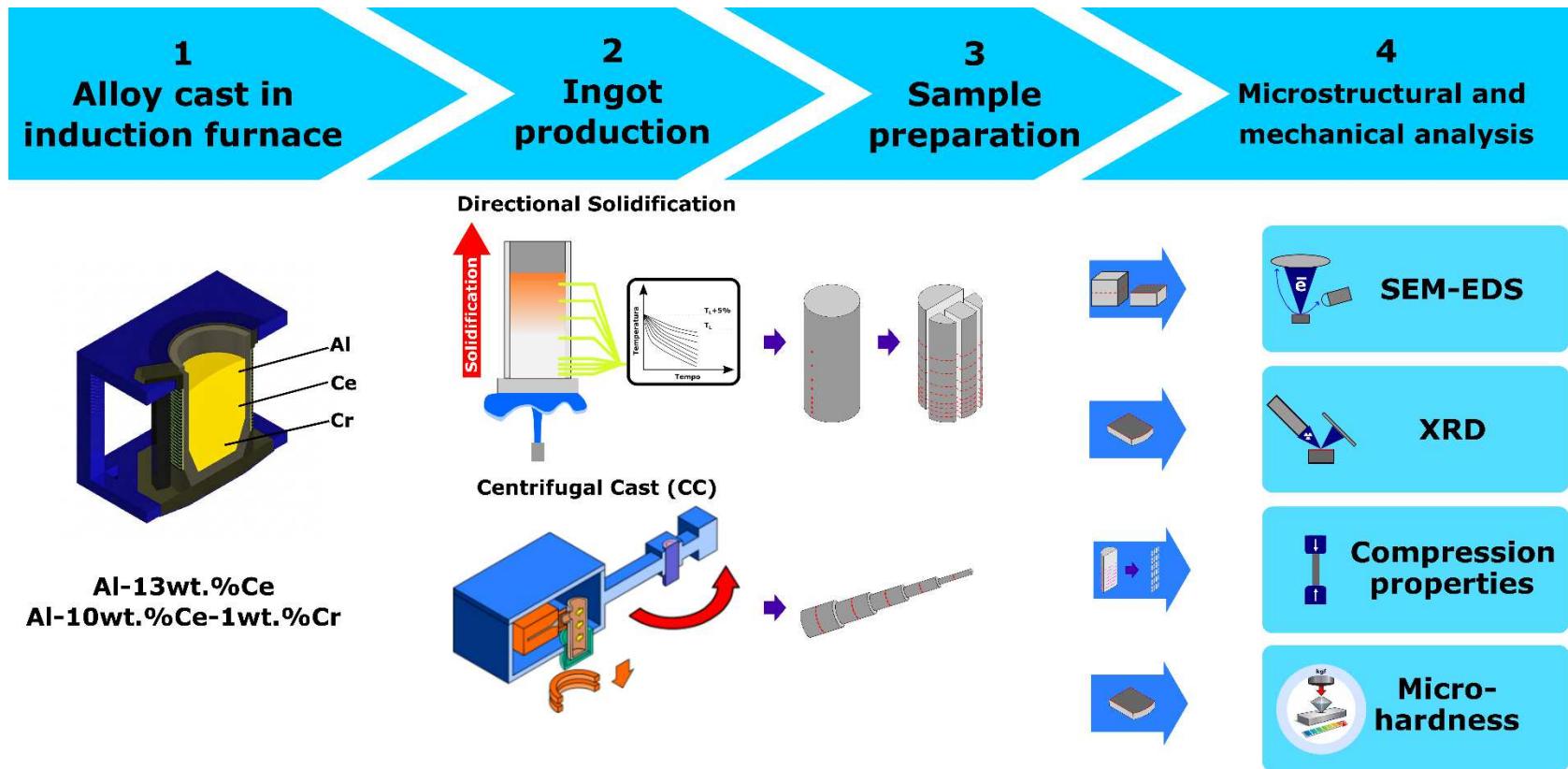


Figure 1 – Schematic representation of the experimental workflow highlighting the directional and centrifugal solidification routes, microstructural analysis, and mechanical characterization via compression testing for Al-13 wt.%Ce and Al-10 wt.%Ce-1 wt.%Cr alloys.

2.2 *Microstructural characterization*

Phase identification was performed by X-ray Diffraction (XRD) using a Bruker D8 Advance diffractometer with Cu-K α radiation, operating in the 2θ range from 5° to 90° . For this work, XRD analyses were conducted exclusively on samples from centrifugal solidification (CS), corresponding to the 5, 7, 9 and 11 mm diameters of the mold cavities, for both Al-13%Ce and Al-10%Ce-1%Cr alloys. The obtained cylinders were sectioned to produce samples with circular profiles of 5 mm diameter and 7 mm length.

For morphological and quantitative analyses, the samples were mounted, ground, polished with 0.05 μm alumina solution, and chemically etched with 5 M NaOH for approximately 10 s to reveal the intermetallic phases. Characterization was performed by scanning electron microscopy (SEM Philips XL30 FEG) in backscattered electron (BSE) mode, allowing observation of α -Al dendrites, eutectic cells, and intermetallic particles, $\text{Al}_{11}\text{Ce}_3$ in the binary alloy and $\text{Al}_{20}\text{CeCr}_2$ in the ternary alloy. Chemical mapping via EDS (dot-mapping) was employed to confirm the partitioning of Ce and Cr between phases. Microstructural quantification included the measurement of secondary dendrite arm spacing (λ_2) for DS samples, average spacing ($\bar{\lambda}$) for CS samples, and eutectic spacing (λ_E) for both DS and CS samples, using the linear intercept method of Heyn [21,22] with the aid of ImageJ software.

2.3 *Mechanical characterization*

The evaluation of the mechanical properties of the Al-Ce and Al-Ce-Cr alloys was designed to establish precise correlations between microstructural refinement and the resulting mechanical response. For mechanical strength characterization, uniaxial compression tests were performed on cylindrical specimens with 5 mm diameter and 7 mm height, which were

extracted from transverse sections of the solidified ingots and machined with the diameter parallel to the growth direction. These tests were conducted on an Instron 5500R universal testing machine, strictly following ASTM E9-19 [23] standard, at a strain rate of $1 \times 10^{-3} \text{ s}^{-1}$ and targeting a total strain of 30%. From the acquired data, engineering and true stress-strain curves were generated, allowing the calculation of plasticity parameters, such as the strain hardening exponent (n) and the strength coefficient (K), through the application of the Hollomon relationship ($\sigma = K \cdot \varepsilon^n$) to evaluate the plastic response of the material.

Additionally, localized resistance was assessed by Vickers hardness testing, using a Shimadzu HMV-G20ST microhardness tester in accordance with ASTM E92-17 [24]. The procedure consisted of applying a load of 4.904 N for a period of 10 s, with 10 indentations performed on each sample to ensure statistical representativeness of the mean values and standard deviations. Measurements were focused on encompassing both the α -Al matrix and the eutectic constituents, allowing the identification of the impact of secondary phase morphology on the final hardness. This integrated methodological approach enabled the correlation of solidification thermal parameters, such as cooling rate, with load-bearing capacity and deformation mechanisms, evidencing how the refinement of intermetallic particles acts as a barrier to dislocation motion.

3 Results and discussions

3.1 *Microstructural comparison of different processing routes*

Figure 2 shows the XRD patterns of the centrifugally solidified Al–13Ce and Al–10Ce–1Cr alloys for the four investigated sample diameters (5, 7, 9, and 11 mm). For the Al–13Ce alloy, the diffraction patterns are dominated by the α -Al phase, with the most intense reflections located at approximately 38° , 45° , 65° , and 78° , together with several lower-intensity peaks

assigned to $\text{Al}_{11}\text{Ce}_3$, particularly in the 2θ ranges of approximately $18\text{--}35^\circ$ and $40\text{--}55^\circ$. The presence of $\text{Al}_{11}\text{Ce}_3$ is consistent with the $\alpha\text{-Al} + \text{Al}_{11}\text{Ce}_3$ eutectic and few primary Ce-rich intermetallic particles as will be seen through SEM in the next results.

For the Al–10Ce–1Cr alloy, $\alpha\text{-Al}$ and $\text{Al}_{11}\text{Ce}_3$ reflections are also detected, while additional peaks associated with the $\text{Al}_{20}\text{CeCr}_2$ phase indicate the formation of the Cr-containing intermetallic. The relative intensities of the $\alpha\text{-Al}$ peaks vary considerably with sample diameter in both alloys, especially for the reflections near $38\text{--}45^\circ$, which is likely associated with differences in preferred crystallographic orientation generated during centrifugal solidification rather than major changes in phase constitution. Overall, the XRD results indicate that changing the casting diameter primarily modifies the microstructural arrangement and crystallographic orientation, while the main formed phases remain essentially unchanged within each alloy.

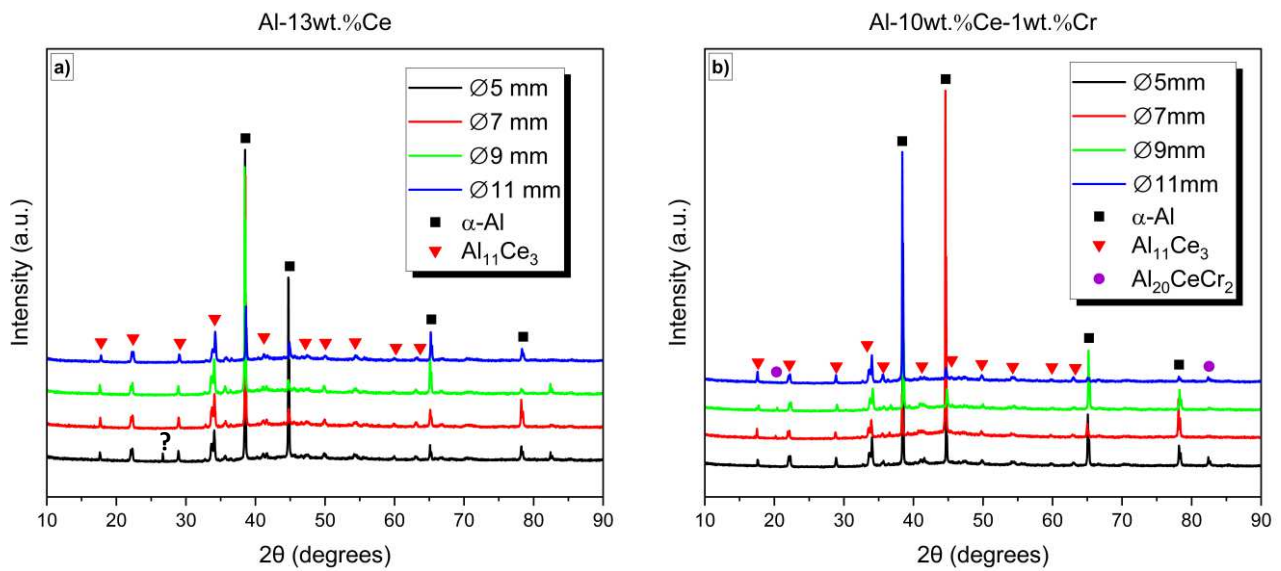


Figure 2 – X-ray diffraction patterns of the centrifugally solidified (CS) (a) Al–13 wt.%Ce and (b) Al–10 wt.%Ce–1 wt.%Cr alloys obtained for sample diameters of 5, 7, 9, and 11 mm.

Figure 3 shows representative SEM micrographs and corresponding EDS dot-mapping of the Al–13 wt.%Ce and Al–10 wt.%Ce–1 wt.%Cr alloys produced by centrifugal cast.

According to the CALPHAD calculations reported by De Paula et al. [20], both compositions are hypereutectic, solidifying through the initial formation of primary intermetallic compounds followed by the eutectic reaction under equilibrium conditions. In the Al–13 wt.%Ce alloy (Figure 3a), the microstructure consists of primary $\text{Al}_{11}\text{Ce}_3$ particles (P) surrounded by a fine eutectic (E) composed of α -Al and $\text{Al}_{11}\text{Ce}_3$ phases, with α -Al (Al) forming the matrix. In the Al–10 wt.%Ce–1 wt.%Cr alloy (Figure 3b), a similar hypereutectic microstructure is observed, consisting of primary $\text{Al}_{20}\text{CeCr}_2$ (P) and an α -Al + $\text{Al}_{11}\text{Ce}_3$ eutectic (E). The EDS dot maps at right in Figure 3 confirm the elemental partitioning, where the red, green, and purple colors correspond to the distribution of Al, Ce, and Cr, respectively. Higher color intensity indicates a higher local X-ray count (higher relative concentration) of the corresponding element, whereas darker regions represent lower signal intensity or the absence of that element. The dot maps corroborate the preferential enrichment of Ce in the primary and eutectic intermetallic phases in the binary alloy as well as the localization of Cr within the primary Cr-containing intermetallic particles in the ternary alloy.

Although the alloys are hypereutectic, the high cooling rates and interface velocities imposed by centrifugal casting promote substantial non-equilibrium undercooling and solute redistribution, allowing α -Al dendritic growth to compete with primary intermetallic and eutectic solidification. The resulting dendritic arrays in Figure 3, therefore, reflect kinetic solidification effects rather than the equilibrium solidification sequence predicted based on the alloy chemistry. Indeed, Ye et al. [25] demonstrated a cooling-rate-induced transition from hypereutectic to hypoeutectic microstructures in a binary Al–14Ce alloy. Above a critical cooling rate determined for their experimental conditions, primary $\text{Al}_{11}\text{Ce}_3$ was replaced by α -Al, as observed here. Their kinetic analysis further showed that, at higher undercoolings, the growth velocity of α -Al exceeds that of $\text{Al}_{11}\text{Ce}_3$, favoring the formation of primary α -Al under rapid solidification conditions.

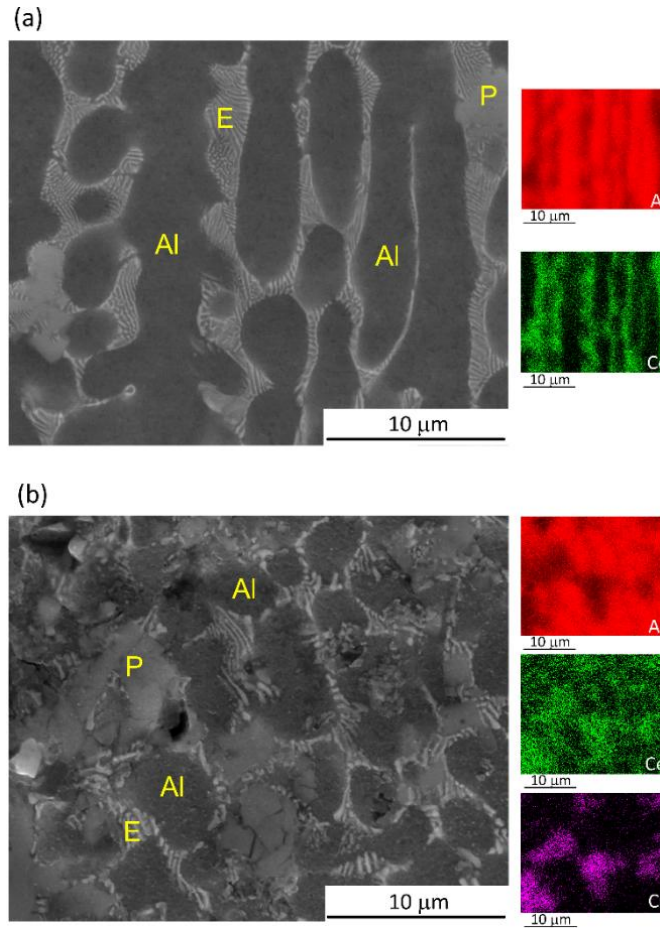


Figure 3 - Representative SEM micrographs and EDS dot maps of the 11-mm-diameter centrifugally cast (a) Al–13Ce and (b) Al–10Ce–1Cr alloys. Al indicates the Al-rich phase, E the eutectic microconstituent, while P denotes the primary formed phase.

Unlike Figure 3, which identifies the constituent phases, Figure 4 highlights the a-Al morphological evolution across the different sections of the copper mold. Therefore, Figure 4 shows the micromorphologies obtained for the centrifugally solidified Al–13 wt.%Ce alloy for sample diameters of (a,b) 5 mm, (c) 7 mm, (d) 9 mm, and (e) 11 mm. The 5 and 7 mm samples exhibit the simultaneous development of two growth morphologies, namely eutectic cells and a-Al dendrites. In contrast, the 9 and 11 mm samples display a predominantly dendritic arrangement, with more elongated and strongly oriented features, indicating pronounced crystallographic texture.

The mixed cellular/dendritic morphology observed in the 5 and 7 mm samples is attributed to the high solid/liquid interface velocities attained during centrifugal casting in copper mold [18]. At these velocities, which are well above the critical velocity for dendrite instability, primary dendrites become extremely fine with suppressed side branching, giving rise to a morphology that closely resembles cellular growth. Consequently, these samples exhibit a transition regime in which cellular and dendritic morphologies coexist. This formation is further enhanced by the alloy composition (Al–13 wt.% Ce), which lies only slightly above the eutectic composition (~10.6 wt.% Ce) [26]. Therefore, only a limited fraction of primary $\text{Al}_{11}\text{Ce}_3$ and dendrites tends to form before the remaining liquid solidifies through the eutectic reaction. Owing to the high eutectic fraction, the microstructure is dominated by cooperative eutectic growth, in which $\alpha\text{-Al}$ and $\text{Al}_{11}\text{Ce}_3$ grow following a highly efficient diffusion-coupled solute redistribution, naturally promoting the formation of eutectic cells. This behavior agrees with Czerwinski and Amirkhiz [26], who showed that even hypereutectic Al–Ce alloys (11–15 wt.% Ce) solidify predominantly through the $\alpha\text{-Al} + \text{Al}_{11}\text{Ce}_3$ eutectic transformation, with only a very limited amount of primary $\text{Al}_{11}\text{Ce}_3$ forming prior to the eutectic reaction. Consequently, the high eutectic fraction seems to facilitate the formation of eutectic cells under rapid solidification conditions. Other related works [25,27] demonstrated that Ce rejection enriches the remaining liquid toward the eutectic composition, allowing the residual melt to solidify as Al/ $\text{Al}_{11}\text{Ce}_3$ eutectic cells (so-called eutectic colonies).

In the 9 and 11 mm samples, the larger volume-to-surface ratio and longer heat-transfer path decrease the local cooling rate and interface velocity, avoiding condition to form cells to be attained. It appears that the more persistent directional heat flow in the larger samples favors competitive growth of favorably oriented grains, resulting in elongated dendrites and a more strongly textured microstructure.

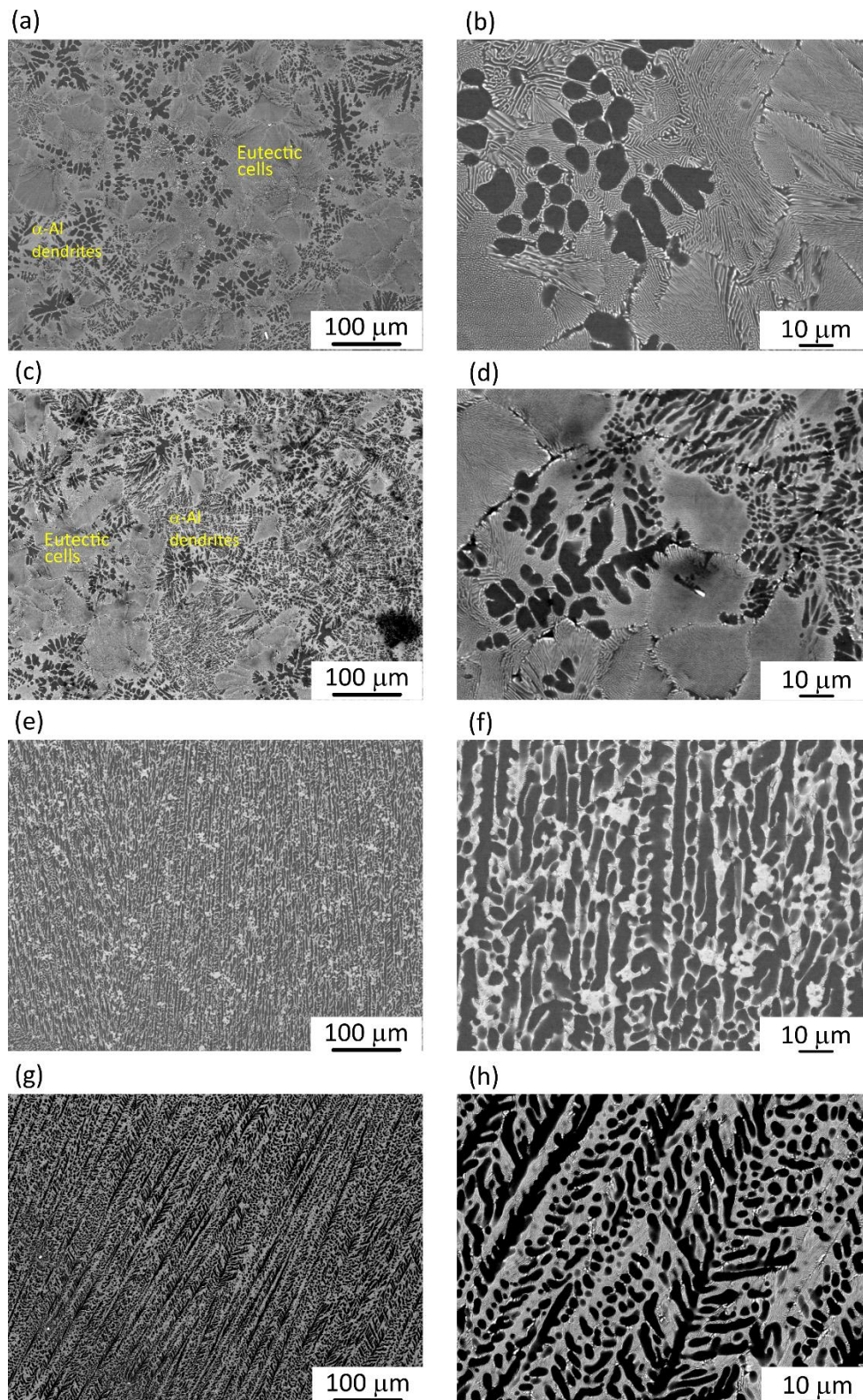


Figure 4 – Representative SEM micrographs of the centrifugally solidified (CS) Al-13wt.%Ce alloy with diameters of 5 mm (a,b), 7 mm (c,d), 9 mm (e,f), and 11 mm (g,h). Left and right columns correspond to low- and high-magnification images, respectively.

Figure 5 depicts the microstructures of the centrifugally cast Al–10 wt.%Ce–1 wt.%Cr alloy obtained from mold diameters of (a,b) 5 mm, (c,d) 7 mm, (e,f) 9 mm, and (g,h) 11 mm. In contrast to the Al–13 wt.%Ce alloy, no cellular (colony) morphology is observed, and a refined dendritic array persists across all mold diameters. This behavior may be associated with the combined effects of the lower Ce content and Cr addition, which modify solute redistribution and promote higher instability of the α -Al/liquid interface, thereby favoring dendritic growth over cellular/eutectic growth. In all conditions, the α -Al dendrites are surrounded by a fine α -Al + $\text{Al}_{11}\text{Ce}_3$ eutectic distributed throughout the interdendritic regions. The 5 and 7 mm samples are characterized by extremely fine dendrites with limited preferential orientation, whereas the 9 mm sample exhibits more elongated and preferentially aligned dendrites, reflecting enhanced competitive growth during solidification. Moreover, the fraction of primary $\text{Al}_{20}\text{CeCr}_2$ intermetallic particles increases with mold diameter, suggesting that the lower cooling rates provide more time for nucleation and growth of the primary intermetallic prior to the α -Al growth and eutectic reaction.

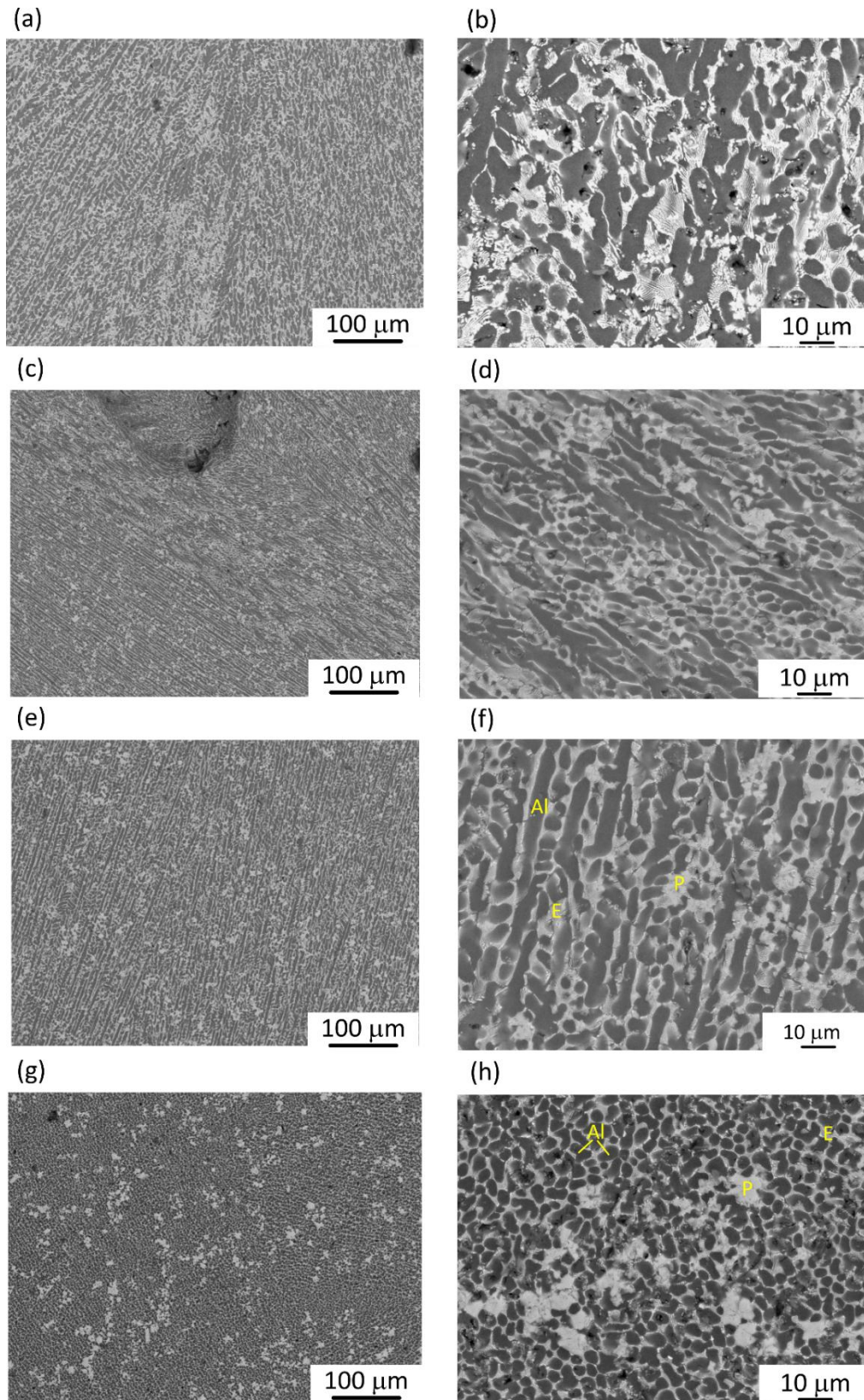


Figure 5 – Representative SEM micrographs of the centrifugally solidified (CS) Al-10wt.%Ce-1wt.%Cr alloy with diameters of 5 mm (a,b), 7 mm (c,d), 9 mm (e,f), and 11 mm (g,h). Left and right columns correspond to low- and high-magnification images, respectively. Al indicates the Al-rich phase, E the eutectic microconstituent, while P denotes the primary formed phase.

Having characterized the microstructures produced by centrifugal casting, directionally solidified (DS) samples were also analyzed to provide a better understanding of the solidification behavior of these alloys. In contrast to centrifugal casting, DS provides a more stable and well-controlled solidification dynamics, characterized by a unidirectional thermal gradient, well-defined growth conditions, and minimal melt convection, allowing a clearer correlation between solidification parameters (i.e., cooling rate) and the resulting microstructure. Figure 6 shows representative SEM micrographs of the DS Al–13 wt.%Ce alloy at cooling rates of (a) 15.8 K/s and (b) 0.7 K/s, and the Al–10 wt.%Ce–1 wt.%Cr alloy at (c) 19.8 K/s and (d) 0.7 K/s. The same microconstituents identified in the CS samples are observed, namely primary α -Al (Al), primary particles (P), and the α -Al + Al₁₁Ce₃ eutectic (E). However, the eutectic exhibits coarser morphology due to the lower solidification velocities and diffusion-controlled coupled growth under more stable thermal conditions allowed through directional solidification.

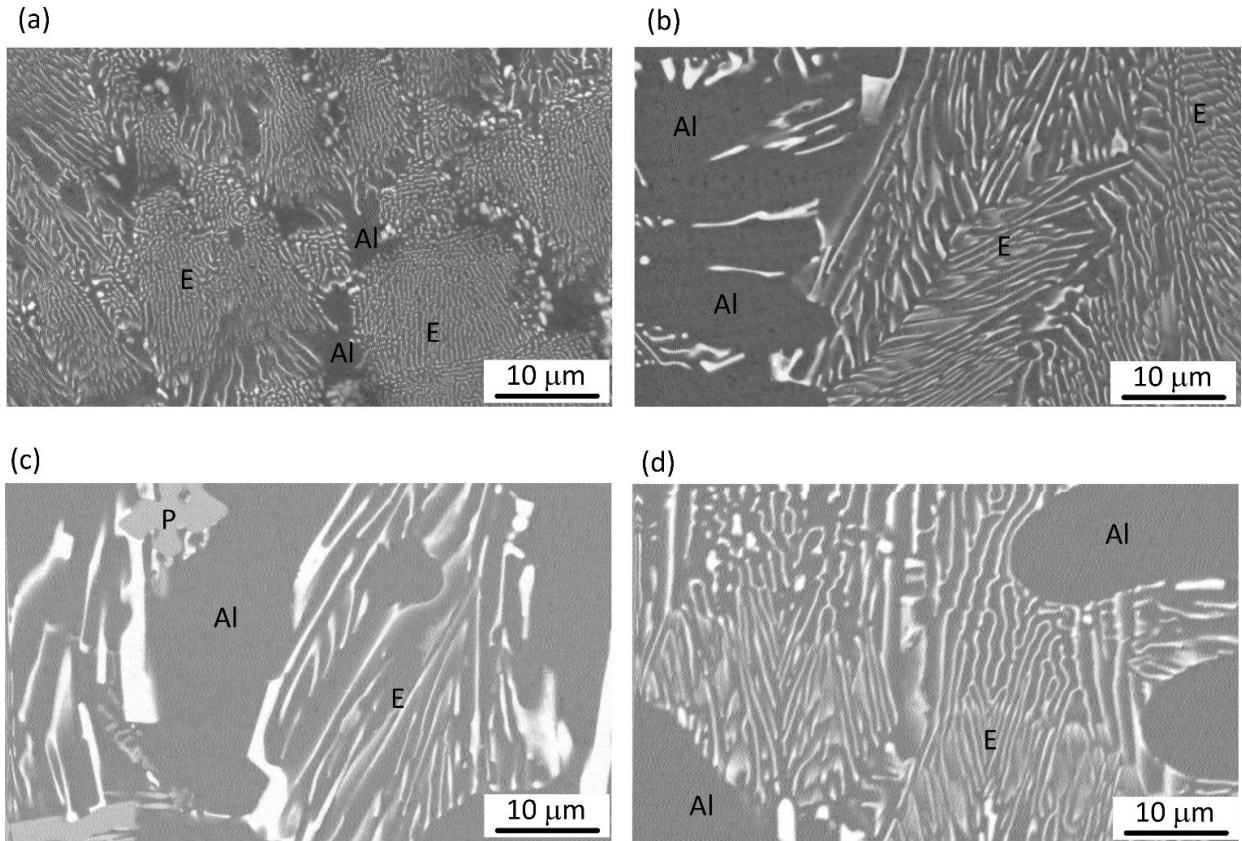


Figure 6 – BSE-SEM micrographs of the directionally solidified (DS) (a,b) Al-13Ce and (c,d) Al-10Ce-1Cr alloys solidified under different cooling rates. Images (a) and (b) correspond to the Al-13Ce alloy processed at cooling rates of 15.8 K/s and 0.7 K/s, respectively, whereas (c) and (d) correspond to the Al-10Ce-1Cr alloy processed at 19.6 K/s and 0.7 K/s, respectively. Al indicates the Al-rich phase and E the eutectic microconstituent, while P denotes the primary AlCeCr phase.

Figure 7(a) shows the eutectic spacing (λ_E) of the directionally solidified samples for the Al-13wt.%Ce and Al-10wt.%Ce-1wt.%Cr alloys. As seen, the ternary alloy exhibited larger eutectic spacings than the Al-Ce alloy under all investigated cooling rates. Furthermore, while the three λ_E data points for the Al-Ce alloy show a tendency for λ_E to decrease with increasing cooling rate, those for the Al-Ce-Cr alloy show a deviation from this expected λ_E refinement trend. This behavior has been previously discussed by De Paula et al. [20] for the same Al-10wt.%Ce-1wt.%Cr alloy, where local variations in the λ_E values were associated with the presence and motion of primary intermetallic particles. These particles were observed to modify the local solidification conditions and, consequently, the eutectic growth behavior.

Figure 7(b) shows that the eutectic spacing (λ_E) in centrifugally solidified samples exhibits a more pronounced variation with sample diameter ($\emptyset$) for the Al-13wt.%Ce alloy than for the Al-10wt.%Ce-1wt.%Cr alloy. For the binary Al-Ce alloy, λ_E decreased from 0.65 μm at $\emptyset 5$ mm to 0.25 μm at $\emptyset 9$ mm, followed by a slight increase to 0.29 μm at $\emptyset 11$ mm. It is worth noting that the $\emptyset 5$ and $\emptyset 7$ mm samples exhibited mixed cellular-dendritic morphologies. In contrast, the Al-Ce-Cr alloy exhibited a smaller variation in λ_E , ranging from 0.27 to 0.43 μm , indicating that the addition of Cr may have reduced the dependence of λ_E on $\emptyset$ and, by association, on the corresponding cooling conditions. Overall, considering the experimental scatter represented by the error bars in Figure 7(b), no pronounced variation in λ_E is observed for either alloy across the four sample diameters. Overall, λ_E can be considered relatively insensitive to sample diameter under the investigated CS conditions.

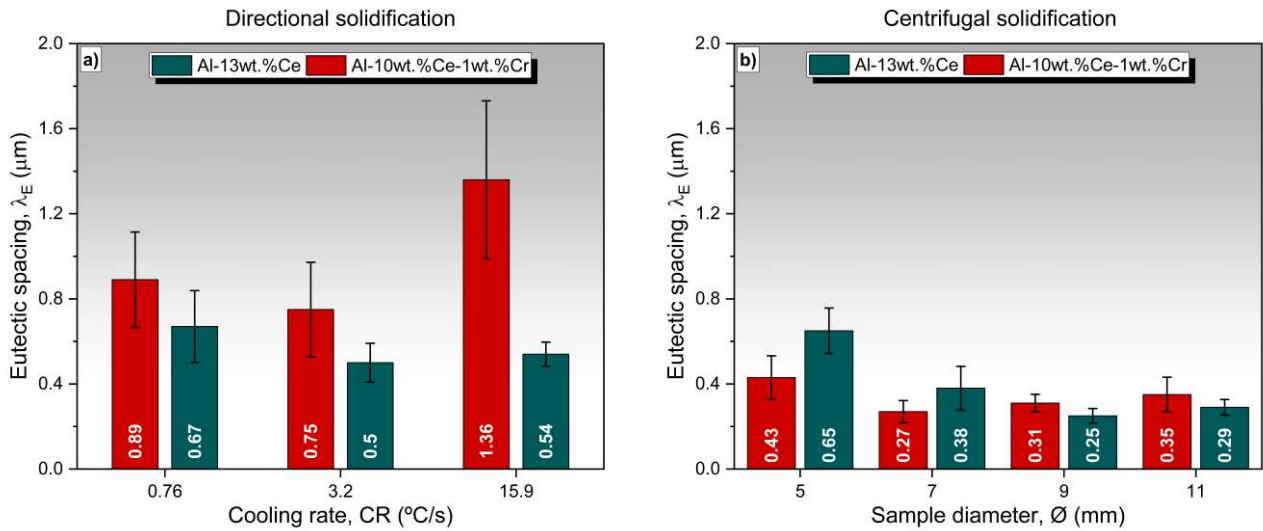


Figure 7 – Eutectic spacing of (a) directionally and (b) centrifugally solidified Al-13Ce and Al-10Ce-1Cr alloys.

3.2 Mechanical properties

Figure 8 shows the compressive stress-strain behavior of the centrifugally cast Al-13 wt.%Ce and Al-10 wt.%Ce-1 wt.%Cr alloys. For the Ø11 mm samples, shown in Figures 8 (a,b) as typical examples of the methodology employed here, the typical engineering (solid lines) and true (dash lines) stress curves remain nearly coincident at low strains but progressively diverge after approximately 0.05 strain because the true stress accounts for the continuous increase in the specimen cross-sectional area during deformation. At a strain of 0.30, the engineering stress typically reaches 260 MPa for the Al-13Ce and 275 MPa for the Al-10Ce-1Cr. Thus, for this sample diameter, the Cr-containing alloy exhibits higher compressive stress throughout most of the plastic deformation range.

Figures 8(c,d) compare the true stress-strain curves for the different casting diameters. For both alloys, the compressive response generally improves with increasing diameter, with the Ø11 and Ø9 mm samples exhibiting the highest stresses, whereas the Ø5 mm condition shows the lowest values. Overall, the Al-10Ce-1Cr alloy samples consistently develop higher true stresses than the Al-13Ce alloy for comparable diameters, indicating that Cr addition improves the compressive strength, while the larger-diameter casting conditions provide the highest flow stresses.

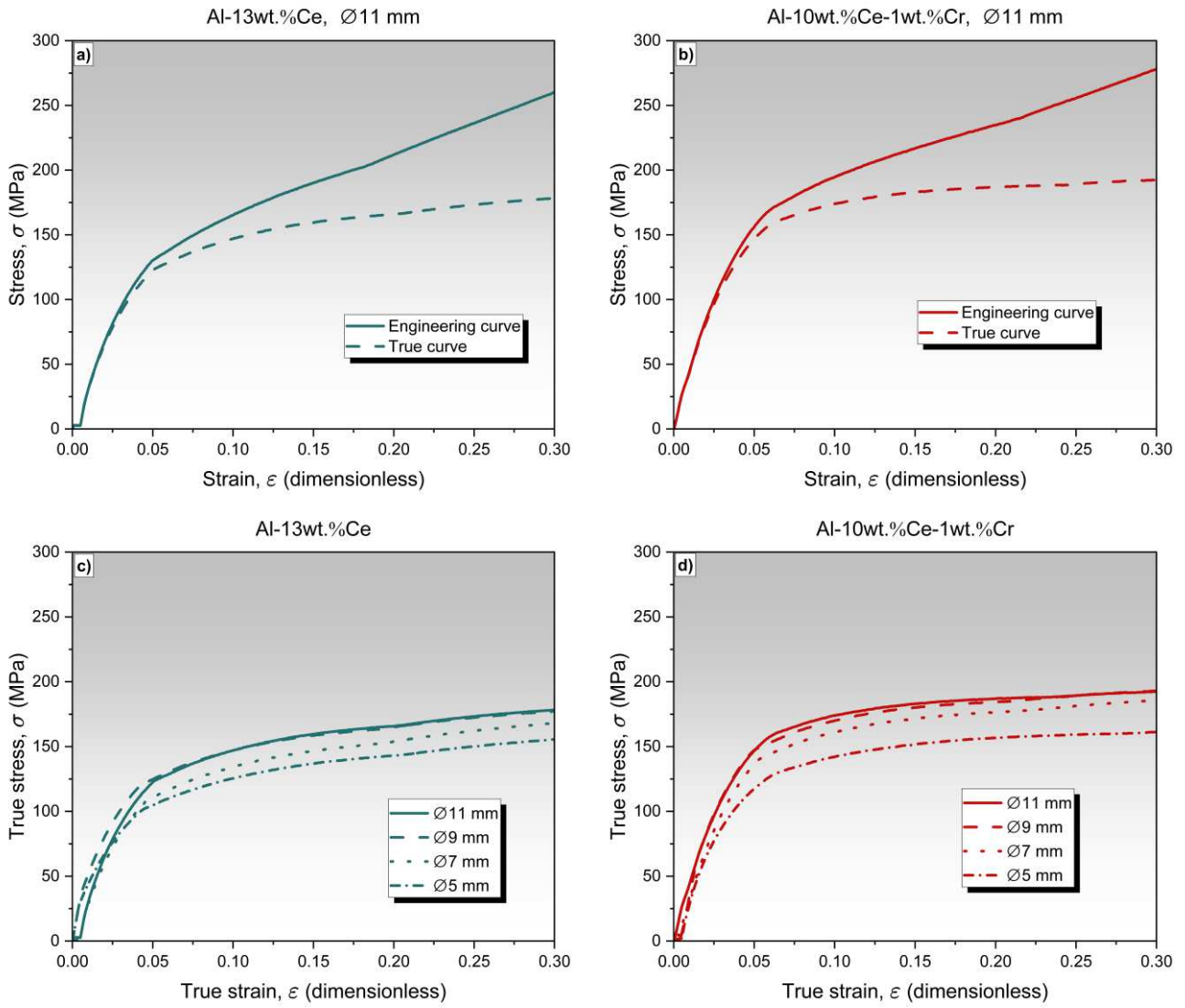


Figure 8 – Typical Engineering and true stress–strain curves for the 11-mm-diameter centrifugally cast (a) Al–13Ce and (b) Al–10Ce–1Cr alloys; and comparison of true stress–strain curves for the CS samples with diameters of 5, 7, 9, and 11 mm in the (c) Al–13Ce and (d) Al–10Ce–1Cr alloys.

Figure 9 and Table 1 compares the engineering compressive strength and Vickers microhardness of the Al–13Ce (green bars) and Al–10Ce–1Cr (red bars) alloys processed by directional solidification (DS) and centrifugal solidification (CS). The corresponding secondary dendritic arm spacing (λ_2) for the DS samples and average dendritic spacing ($\bar{\lambda}$) for the CS samples are indicated inside each bar. The compressive strength and Vickers hardness values for the 5 and 7 mm Al–13Ce alloy samples have not been included in Figure 9, since

these sections correspond to a cellular-to-dendritic transition (CDT) region and therefore cannot be directly correlated with a representative dendritic spacing.

Regarding the microstructural spacings indicated inside the bars, centrifugal solidification produces substantially finer structures than directional solidification. For the DS samples, λ_2 varies from approximately 5.0 to 7.8 μm for the Al–13Ce alloy and from 16.7 to 23.6 μm for the Al–10Ce–1Cr alloy, whereas the CS samples exhibit average dendritic spacings of 4.0–5.4 μm ($\bar{\lambda}$ measured only for the 9 and 11 mm samples, as the 5 and 7 mm sections exhibited mixed cellular–dendritic morphologies in the Al–13Ce alloy). This pronounced refinement is primarily associated with the more intense heat extraction and higher solidification rates imposed by the metallic mold during centrifugal casting. In addition, forced melt flow and convection during CS enhance heat and solute transport, further affecting the stability and growth kinetics of the solidification front. The Al–13Ce alloy consistently exhibits smaller characteristic spacings than the Al–10Ce–1Cr alloy under both directional solidification and centrifugal casting. This behavior can be primarily associated with its higher Ce content, which enhances solute redistribution and constitutional undercooling ahead of the $\alpha\text{-Al}$ /liquid interface, restricting lateral dendritic growth and promoting a finer microstructural scale. The larger eutectic fraction of the binary alloy may additionally constrain the growth of the primary $\alpha\text{-Al}$ dendrites.

Under DS, the Cr-containing alloy consistently exhibits higher compressive strength as compared to the DS binary alloy. This improvement can be associated with the microstructural modifications promoted by Cr addition, particularly the formation of primary $\text{Al}_{20}\text{CeCr}_2$ particles and the persistence of a dendritic array, in contrast to the higher contribution of eutectic regions in the binary alloy. Although coarse, the hard $\text{Al}_{20}\text{CeCr}_2$ particles can act as effective barriers to dislocation motion under compression load, promoting dislocation accumulation and increasing the resistance to plastic flow. Moreover, unlike under tensile

loading, where coarse brittle intermetallics may favor crack initiation and premature failure, their detrimental effect is less pronounced under compression, allowing their reinforcing contribution to prevail [28,29]. It can be observed that the compressive strength does not exhibit a simple dependence on λ_2 , indicating that the morphology, size, fraction, and distribution of the constituent phases are at least as important as the dendritic scale in controlling the macroscopic mechanical response.

For CS, both alloys exhibit slightly higher compressive strengths, with this behavior being more evident for the binary alloy in Figure 9. The Al–10Ce–1Cr alloy generally retains the higher compressive strength, although the difference between the alloys becomes smaller and some scatter is observed. The much finer spacing generated by centrifugal casting does not necessarily translate into a proportional increase in compressive strength, since the resulting phase morphology and distribution are also substantially modified.

Interestingly, Vickers hardness in Figure 9(c,d) does not follow the compressive-strength trend. Under DS, the binary Al–13Ce alloy exhibits higher hardness values than the Al–10Ce–1Cr alloy, despite its lower compressive strength. The lower hardness of the Cr-containing alloy can be primarily associated with its coarser eutectic constituent, whereas the relatively low load employed during hardness testing limits the contribution of the comparatively low-fraction primary $\text{Al}_{20}\text{CeCr}_2$ particles to the measured values. Microhardness probes a small and highly localized volume and is therefore strongly influenced by whether the indentation intercepts the α -Al matrix, eutectic constituent, or an intermetallic particle. Compression, in contrast, reflects the collective response of the entire multiphase microstructure, including second phase–matrix interactions, load transfer, phase connectivity, and resistance to plastic flow. For the CS samples, the hardness values of the two alloys are considerably closer, consistent with their finer and more homogeneous microstructures.

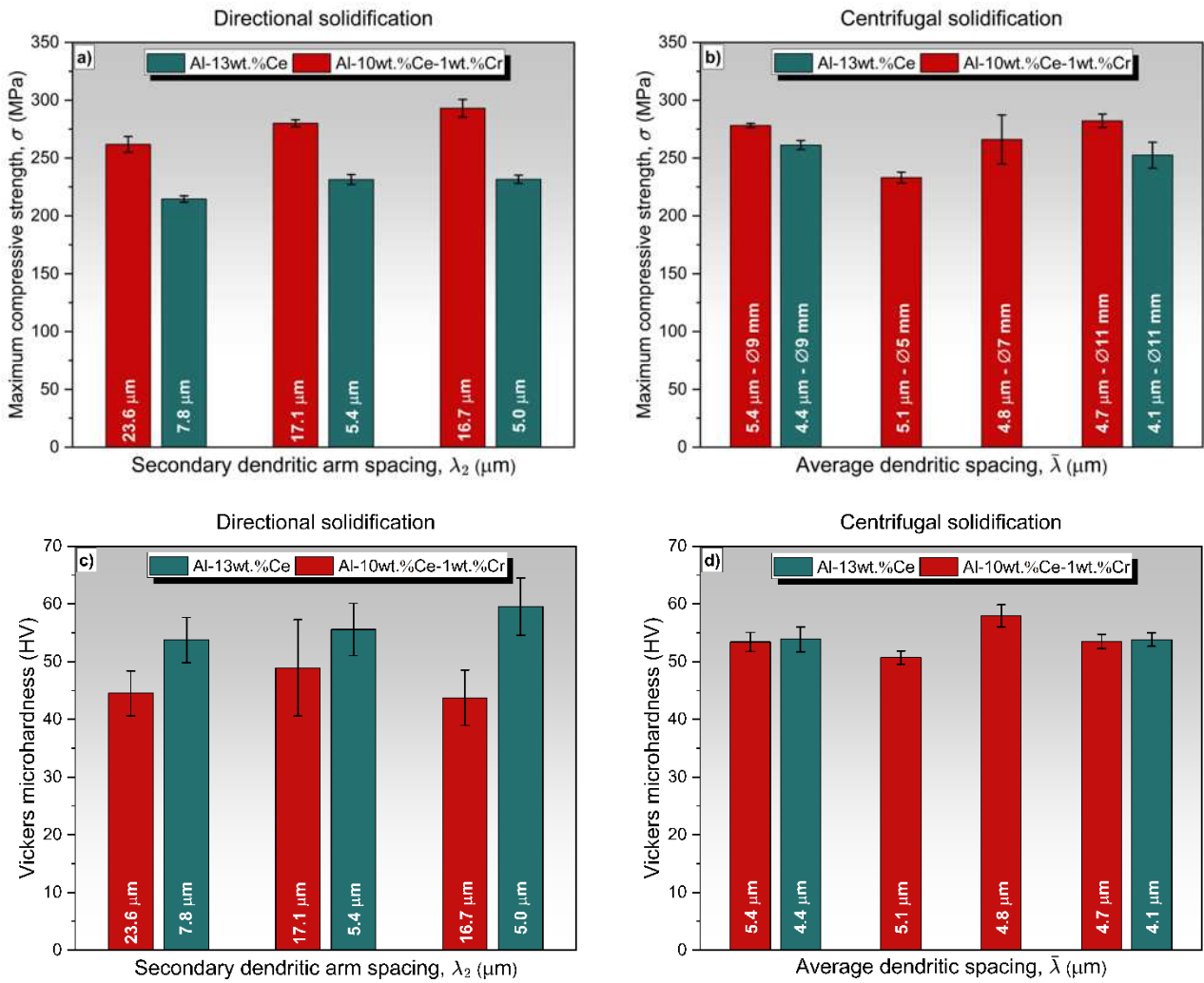


Figure 9 - Relationship between maximum engineering compressive strength and Vickers hardness as a function of the secondary dendritic arm spacing (λ_2) for directionally solidified alloys (a,c) and the average dendritic spacing ($\bar{\lambda}$) for centrifugally cast alloys (b,d). Compressive strength: (a,b). Vickers microhardness: (c,d).

The Al-10Ce-1Cr alloy exhibits consistently higher yield strength in Figure 10 and Table 1, particularly after centrifugal solidification. Since both alloys exhibit similarly fine dendritic spacings under CS, this improvement cannot be attributed primarily to microstructural refinement. Instead, the rigid primary $\text{Al}_{20}\text{CeCr}_2$ particles restrict plastic deformation and promote dislocation accumulation at the particle/matrix interfaces, increasing the stress required for yielding. Thus, phase constitution and particle-matrix interactions appear to dominate in controlling yield strength.

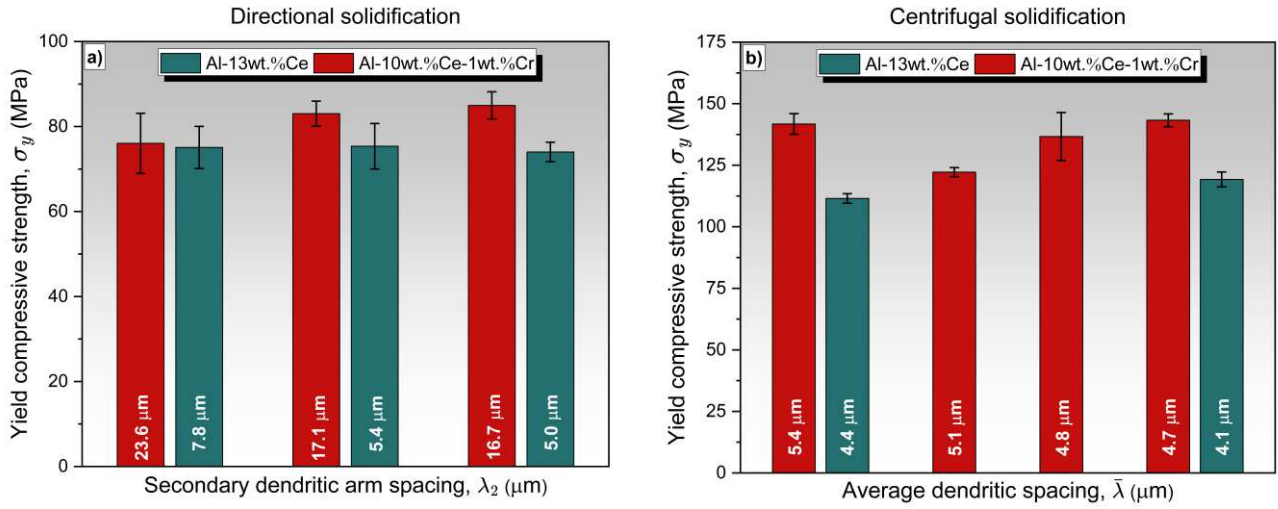


Figure 10 – Yield engineering compressive strength of the Al–13Ce and Al–10Ce–1Cr alloys as a function of the secondary dendritic arm spacing (λ_2) for directionally solidified samples (a) and the average dendritic spacing ($\bar{\lambda}$) for centrifugally cast samples (b).

Table 1 summarizes the mechanical and microstructural parameters of the Al–13Ce and Al–10Ce–1Cr alloys, including engineering and true compressive properties, the strain-hardening exponent (n), strength coefficient (K), characteristic microstructural spacing (λ), and Vickers microhardness. The parameters n and K , derived from the Hollomon relationship, provide complementary measures of the plastic response: n describes the strain-hardening capability, whereas K represents the characteristic stress level associated with plastic flow. In cast Al alloys, n is particularly sensitive to microstructural features such as secondary dendritic spacing and particle morphology and is therefore useful for assessing the relationship between solidification microstructure and plastic deformation behavior [30].

Considering the overall average values considering both DS and CS, the Al–13Ce alloy exhibited $n = 0.18$ and $K = 205$ MPa, whereas the Al–10Ce–1Cr alloy showed a lower strain-hardening exponent ($n = 0.14$) but a higher strength coefficient ($K = 231$ MPa). This indicates that the Cr-containing alloy offers higher resistance to plastic flow, while the binary alloy exhibits a slightly higher ability to progressively strain harden during deformation. The higher

“K” of the Cr-containing alloy is consistent with the higher fraction of rigid $\text{Al}_{20}\text{CeCr}_2$ particles as can be seen in Figure 5, which constrain matrix plasticity and increase the stress required for deformation. Conversely, the lower “n” suggests a reduced capacity for additional dislocation storage and multiplication as deformation proceeds, likely because the hard intermetallic-reinforced microstructure already imposes strong barriers to dislocation motion. Accordingly, the binary alloy starts from a lower flow-stress level but develops a comparatively stronger strain-hardening response.

Table 1 – Summary of the compressive properties and microstructural parameters of the Al–13Ce and Al–10Ce–1Cr alloys. CDT refers to cellular-to dendritic transition regions.

Alloy	Process	CR/Ø	σ (MPa)	σ_y (MPa)	E (MPa)	U_t (MPa)	$\sigma_{30\%,true}$ (MPa)	$\sigma_{y,true}$ (MPa)	E_{true} (MPa)	$U_{t,true}$ (MPa)	n	K (MPa)	λ (μm)	HV	λ_E (μm)
Al-13wt.%Ce	DS	15.9 °C/s	231.7 ± 3.6	74.0 ± 2.3	2997.1 ± 294.5	49.3 ± 1.3	162.8 ± 2.4	71.8 ± 2.5	2861.6 ± 290.2	49.3 ± 1.3	0.184 ± 0.010	213.1 ± 2.3	5.04 ± 0.69	59.53 ± 4.96	0.538 ± 0.057
		3.2 °C/s	231.5 ± 4.3	75.4 ± 5.4	2883.6 ± 249.4	50.2 ± 1.4	164.9 ± 3.7	73.2 ± 5.0	2790.6 ± 75.6	50.2 ± 1.4	0.180 ± 0.001	214.4 ± 7.0	5.44 ± 0.55	55.57 ± 4.53	0.503 ± 0.091
		0.76 °C/s	214.6 ± 2.8	75.1 ± 5.0	2683.3 ± 70.9	44.2 ± 1.1	147.7 ± 2.7	72.0 ± 5.3	2582.3 ± 113.1	44.2 ± 1.1	0.190 ± 0.023	192.7 ± 1.9	7.18 ± 0.53	53.72 ± 3.92	0.669 ± 0.169
	CS	11 mm	252.5 ± 11.2	119.2 ± 3.0	3143.8 ± 496.2	53.3 ± 2.4	173.5 ± 7.8	112.8 ± 4.0	3056.4 ± 462.9	53.3 ± 2.4	0.144 ± 0.025	208.0 ± 10.8	4.06 ± 0.13	53.82 ± 1.18	0.287 ± 0.037
		9 mm	261.3 ± 3.9	111.5 ± 1.9	3209.2 ± 179.3	54.4 ± 0.9	178.9 ± 2.6	107.3 ± 2.2	3023.3 ± 219.5	54.4 ± 0.9	0.167 ± 0.010	217.9 ± 4.5	4.40 ± 0.14	53.86 ± 2.14	0.246 ± 0.035
		7 mm	242.8 ± 5.7	104.6 ± 2.4	3311.0 ± 1012.8	49.9 ± 1.3	165.8 ± 3.6	98.9 ± 1.7	3146.3 ± 912.1	49.9 ± 1.3	0.176 ± 0.011	203.8 ± 4.2	CDT	53.56 ± 2.64	0.379 ± 0.103
		5 mm	225.0 ± 19.8	100.6 ± 13.8	3378.7 ± 1611.3	47.2 ± 4.1	153.6 ± 13.7	95.8 ± 14.0	3867.8 ± 2598.2	47.2 ± 4.1	0.200 ± 0.025	185.9 ± 12.2	CDT	52.89 ± 2.86	0.650 ± 0.107
Al-10wt.%Ce-1wt.%Cr	DS	15.9 °C/s	293.1 ± 7.6	85.0 ± 3.2	3693.0 ± 75.3	62.0 ± 0.6	202.8 ± 5.4	82.5 ± 3.6	3469.1 ± 144.3	62.0 ± 0.6	0.139 ± 0.006	250.2 ± 6.5	16.68 ± 0.69	44.51 ± 3.91	1.362 ± 0.371
		3.2 °C/s	280.1 ± 3.1	83.0 ± 2.9	3410.2 ± 234.1	58.9 ± 1.2	192.9 ± 2.7	80.4 ± 2.8	3240.8 ± 265.7	58.9 ± 1.2	0.182 ± 0.019	256.8 ± 4.4	17.12 ± 1.08	48.94 ± 8.34	0.747 ± 0.222
		0.76 °C/s	261.8 ± 6.8	76.0 ± 7.1	3400.5 ± 121.6	55.6 ± 1.3	181.7 ± 3.4	74.2 ± 6.7	3252.5 ± 117.2	55.6 ± 1.3	0.201 ± 0.012	249.6 ± 0.9	23.58 ± 3.13	43.75 ± 4.79	0.888 ± 0.224
	CS	11 mm	282.2 ± 5.8	143.3 ± 2.6	3559.2 ± 172.0	61.4 ± 0.8	195.4 ± 4.1	139.7 ± 2.5	3220.3 ± 160.5	61.4 ± 0.8	0.107 ± 0.005	225.5 ± 5.2	4.68 ± 0.11	53.51 ± 1.25	0.348 ± 0.081
		9 mm	278.1 ± 1.9	141.8 ± 4.2	3513.2 ± 251.5	59.8 ± 0.7	192.6 ± 1.6	134.2 ± 3.0	3278.7 ± 282.1	59.8 ± 0.7	0.118 ± 0.004	224.7 ± 1.6	5.42 ± 0.23	53.42 ± 1.65	0.309 ± 0.041
		7 mm	266.1 ± 21.2	136.6 ± 9.7	4126.6 ± 2037.8	57.6 ± 5.1	183.5 ± 15.0	130.0 ± 10.8	3920.9 ± 2050.1	57.6 ± 5.1	0.116 ± 0.024	212.5 ± 9.4	4.76 ± 0.20	57.94 ± 1.90	0.266 ± 0.520
		5 mm	233.1 ± 4.6	122.2 ± 1.8	2452.6 ± 324.5	50.3 ± 1.4	160.4 ± 3.0	111.2 ± 7.3	2382.2 ± 110.4	50.3 ± 1.4	0.139 ± 0.009	195.5 ± 1.7	5.15 ± 0.25	50.69 ± 1.15	0.431 ± 0.103

4 Conclusions

The following conclusions can be drawn from the present findings:

- Centrifugal solidification promoted microstructural refinement compared with directional solidification, producing characteristic dendritic spacings of 4.0 to 5.4 μm .
- A cellular-to-dendritic transition was identified in the CS Al–13Ce alloy, with mixed cellular/dendritic morphologies at diameters of 5–7 mm and predominantly dendritic structures at 9–11 mm. This transition is attributed to the higher solidification velocities in the smaller sections and the high eutectic fraction of the Al–13Ce alloy, which favor coupled eutectic cellular growth. In contrast, the lower Ce content combined with Cr addition stabilized dendritic growth in the Al–10Ce–1Cr alloy across all investigated diameters.
- Cr addition significantly altered the solidification behavior and phase constitution, promoting the formation of primary $\text{Al}_{20}\text{CeCr}_2$ intermetallic particles and suppressing the cellular morphology observed in the binary alloy under the highest solidification rates (i.e., smallest diameters).
- The Al–10Ce–1Cr alloy generally exhibited superior compressive performance, particularly in terms of yield strength. Under centrifugal solidification, the highest yield strength reached approximately 143 MPa, compared with 119 MPa for the Al–13Ce alloy, indicating that the $\text{Al}_{20}\text{CeCr}_2$ particles effectively increase resistance to plastic deformation.
- Microstructural refinement alone did not govern the mechanical response. The combined effects of phase constitution, intermetallic distribution, and particle/matrix interactions were decisive for compressive behavior. Consistently, the Al–10Ce–1Cr alloy exhibited a higher average strength coefficient ($K = 231 \text{ MPa}$) but a lower strain-hardening exponent ($n = 0.14$)

than the Al–13Ce alloy ($K = 205$ MPa; $n = 0.18$), demonstrating greater resistance to plastic flow but lower strain-hardening capacity.

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Conflict of interest

The authors declare that they have no conflict of interest.

Data Availability Statement

Data presented in this study are available on request from the corresponding author.

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List of Figure Captions

Figure 1 – Schematic representation of the experimental workflow highlighting the directional and centrifugal solidification routes, microstructural analysis, and mechanical characterization via compression testing for Al-13 wt.%Ce and Al-10 wt.%Ce-1 wt.%Cr alloys.

Figure 2 – X-ray diffraction patterns of the centrifugally solidified (CS) (a) Al-13 wt.%Ce and (b) Al-10 wt.%Ce-1 wt.%Cr alloys obtained for sample diameters of 5, 7, 9, and 11 mm.

Figure 3 - Representative SEM micrographs and EDS dot maps of the 11-mm-diameter centrifugally cast (a) Al-13Ce and (b) Al-10Ce-1Cr alloys. Al indicates the Al-rich phase, E the eutectic microconstituent, while P denotes the primary formed phase.

Figure 4 – Representative SEM micrographs of the centrifugally solidified (CS) Al-13wt.%Ce alloy with diameters of 5 mm (a,b), 7 mm (c,d), 9 mm (e,f), and 11 mm (g,h). Left and right columns correspond to low- and high-magnification images, respectively.

Figure 5 – Representative SEM micrographs of the centrifugally solidified (CS) Al-10wt.%Ce-1wt.%Cr alloy with diameters of 5 mm (a,b), 7 mm (c,d), 9 mm (e,f), and 11 mm (g,h). Left and right columns correspond to low- and high-magnification images, respectively. Al indicates the Al-rich phase, E the eutectic microconstituent, while P denotes the primary formed phase.

Figure 6 – BSE-SEM micrographs of the directionally solidified (DS) (a,b) Al-13Ce and (c,d) Al-10Ce-1Cr alloys solidified under different cooling rates. Images (a) and (b) correspond to the Al-13Ce alloy processed at cooling rates of 15.8 K/s and 0.7 K/s, respectively, whereas (c) and (d) correspond to the Al-10Ce-1Cr alloy processed at 19.6 K/s and 0.7 K/s, respectively. Al indicates the Al-rich phase and E the eutectic microconstituent, while P denotes the primary AlCeCr phase.

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Figure 9 - Relationship between maximum engineering compressive strength and Vickers hardness as a function of the secondary dendritic arm spacing (λ_2) for directionally solidified alloys (a,c) and the average dendritic spacing ($\bar{\lambda}$) for centrifugally cast alloys (b,d). Compressive strength: (a,b). Vickers microhardness: (c,d).

Figure 10 – Yield engineering compressive strength of the Al-13Ce and Al-10Ce-1Cr alloys as a function of the secondary dendritic arm spacing (λ_2) for directionally solidified samples (a) and the average dendritic spacing ($\bar{\lambda}$) for centrifugally cast samples (b).

Table heading

Table 1 – Summary of the compressive properties and microstructural parameters of the Al-13Ce and Al-10Ce-1Cr alloys. CDT refers to cellular-to dendritic transition regions.