Laser Remelting of AlSi10Mg(-Ni) Alloy Surfaces: Influence of Ni Content and Cooling Rate on the Microstructure

Danusa Araújo de Moura
Guilherme Lisboa de Gouveia
Gustavo Figueira
Amauri Garcia
Piter Gargarella
José Eduardo Spinelli (✉ spinelli@ufscar.br)
UFSCar <https://orcid.org/0000-0003-0611-1038>

Research Article

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Abstract

AlSi10Mg alloys are widely employed in a variety of industries, including aerospace, automotive, and microelectronics. This is because of its low density, acceptable mechanical properties, acceptable corrosion resistance, and inexpensive application cost. Advantageous fluidity, a short solidification period, and minimal volumetric contraction are beneficial characteristics under processing such alloys. Despite being used as commercial alloys, the mechanical properties of the AlSi10Mg alloys still need to be improved. In line with this, the current focus of Al-based alloys development is mostly on modifying commercially available alloys. Under such context, Ni was used as an alloying element in this study to generate the Al₃Ni intermetallics, distinguished by its improved mechanical strength. Furthermore, the thermal stability of the Al₃Ni may be a benefit, particularly for high-temperature applications. The present study aims to investigate the solidification under low and high cooling rates of four alloys: AlSi10Mg, AlSi10Mg-1Ni, AlSi10Mg-2Ni, and AlSi10Mg-3Ni (wt.%). Samples were obtained by directional solidification (DS) and laser surface remelting (LSR) processes. The cooling rates were calculated for the DS samples and with extrapolation for LSR samples as well as with the use of a model from the literature. After testing several laser conditions, the results also include an examination of microstructural and hardness changes in the treated and untreated zones. The produced gradient of microstructures is fully characterized as well as used to evaluate cooling rates inside the laser molten pools. For energy densities of 400 J/mm² and 100 J/mm², the mean dendritic spacings, λ , of the three Ni-containing alloys at the laser molten pool yielded estimated cooling rates of approximately $1.5 \cdot 10^4$ °C/s and $4.7 \cdot 10^4$ °C/s, respectively. A model explaining the reversion of λ across the molten pool will be outlined.

1. Introduction

Laser surface treatments are an effective ways to change mechanical and chemical characteristics locally. Laser applications are particularly competitive in the industrial context because of the laser processes' cleanliness, high speed, and automation [1, 2]. The laser surface remelting (LSR) process generates a thin molten region on the material surface by rapidly passing a continuous, high-energy-density laser beam over it. Once the energy source is withdrawn, the capability to keep a cold substrate while remelting a thin surface region results in rapid quenching of the molten region. Extremely high solidification cooling rates (from 10^4 to 10^7 °C/s) have been reported [3]. Moreover, LSR has been proven to produce a wide range of unique and intriguing surface microstructures and features [4–6]. LSR can produce a surface layer with strong metallurgical bonding to the substrate under rapid melting and solidification conditions. Significant grain refinement and microstructure uniformity with fine dendrites/cells in the surface layer can be achieved using this process [7]. Nowadays, there's much interest in Al alloys for pistons, manifolds, gear units, crankcases, and chassis, all constructed mainly of light alloys like Al- and Mg-based ones. Although the density of these materials allows for significant weight and fuel savings, in most situations, it is required to adjust the surface microstructural properties, which may be accomplished using laser surface treatments.

Several variables may influence the interactions between a laser beam and an alloy surface [8]. Within the range of operating conditions, a specific combination of power density and residence time produces a particular operational regime, resulting in unique effects on the material's surface. Physical measurements of crucial parameters such as temperature and velocity fields in the molten pools are also difficult due to the small area of the laser-treated zones and the short time frames [9].

Due to its capacity to produce complicated geometry components in a near-net-shape fashion, several Additive Manufacturing (AM) processes have emerged in the last years [10, 11]. Even though this technology has several processing particularities, the components generated must solidify at high cooling rates (T) [12, 13] due to the use of laser sources in various conditions. High cooling rates in the AM processes have long been known to optimize the microstructure in terms of grain size and dendrite arm spacing (DAS). Owing to that, the Laser Surface Remelting (LSR) technique can aid in the examination of preliminary data before the AM processing. Moreover, transient directional solidification [14, 15] is another key approach for rationalizing cooling rate-microstructure dependences in Al alloys.

It is well known that the cooling rates in the laser molten pool are variable at various points and difficult to detect. As a result, the formation of a gradient microstructure at the surface is often reported [16], being their features and benefits still unclear.

Due to the high Si and proper Mg contents allowing hardening mechanisms to be activated, AlSi10Mg alloys are of major economic interest. These alloys have demonstrated great potential and versatility in several conventional processes involving melting and solidification, such as high pressure die-casting (HPD), permanent mold casting (PM), and lost wax (LW) [17, 18]. Studies on solidification through HPD demonstrated the formation of refined microstructures containing the α -Al dendrites enveloped by the (Al + Si) eutectic for the AlSi10Mg0.6 alloy. In contrast, the PM low-cooling rate process produced the α -Al matrix with a coarse dendritic arrangement and interdendritic regions constituted by the α -Al, Si, and Mg₂Si phases within the eutectic mixture [17, 18].

Moreover, AM has extensively explored and manufactured commercial AlSi10Mg alloy parts [19, 20]. This is one of the few Al alloys that have been successfully employed in AM. Under AM conditions, it evolved to a completely different solidification behavior resulting in very refined microstructures. Yan et al. [21] demonstrated that the microstructure of the additive manufactured AlSi10Mg sample was composed of a thin cellular structure having α -Al cells smaller than 1 μ m in size and with eutectic Si arranged in the intercellular regions. This was mainly because laser melting involves highly concentrated energy input, leading to large local-temperature gradients during solidification. It is worth mentioning that there is a current interest in developing alternative Al-based alloys for the aforementioned processes, which is mostly focused on the modification of commercially available alloys [22, 23].

Under such context, Garmendia et al. [24] added 1 wt.% Cu to the AlSi10Mg alloy. Compared to the AlSi10Mg, the ultimate tensile strength of the Cu-added specimens increased by 51 MPa. This was attributed to grain size reduction and subsequent Hall–Petch strengthening and to Orowan strengthening

induced by Cu-rich particles. Like Cu, the addition of Ni appears to be a promising possibility considering that Al_3Ni particles may be generated, improving mechanical properties.

According to the best of the authors' knowledge, Ni has not yet been attempted in AlSi10Mg alloys. High mechanical strength is expected due to the hard Al_3Ni intermetallics. Furthermore, below 500°C , such intermetallics are proved to be thermally stable [25–27]. Cheung et al. [28] investigated the solidification microstructure of the Al-5Ni alloy remelted by laser surface treatment. In the laser track cross-section, three zones of refined microstructures were revealed: a lamellar eutectic at the bottom of the treated pool, an irregular eutectic pattern, and a dendritic pattern with different growth orientations that culminates in dendritic growth parallel to the laser beam speed direction at the top of the pool.

The isolated reinforcing roles of plate-like Si particles and Al_3Ni fibers for binary Al-11Si and Al-5Ni alloys, respectively, have been determined by Kakitani et al. [29]. For each case, a similar performance in the blockage of dislocations has been demonstrated without any influence of the scale of the dendritic Al-rich matrix (secondary dendritic spacing ranging from 5 to $30\text{ }\mu\text{m}$), resulting in a constant hardness of approximately 48 HV. The simultaneous interaction of both strengthening phases embedded in the interdendritic regions and the scale of the secondary dendritic arms (from 5 to $10\text{ }\mu\text{m}$) was favorable to a notable improvement in hardness for the ternary Al-11Si-5Ni alloy.

The current research focuses on characterizing AlSiMg10(Ni) alloys after laser remelting. To validate the extrapolation technique of cooling rates from DS, an experimental examination was previously carried out under slow solidification. This research also looks at the relationship between laser process parameters, layers generated inside the molten pool, characterization of microstructural spacings, and hardness changes in laser-treated and unmolten zones.

2. Experimental Procedure

The AlSi10Mg , AlSi10Mg-1Ni , AlSi10Mg-2Ni , and AlSi10Mg-3Ni alloys were directionally solidified in cylindrical bipartite molds coated internally with a refractory material to minimize radial heat losses. K-type thermocouples were distributed along with the height of the casting from the water-cooled bottom part to obtain data on temperature either as a function of time (acquisition frequency of 5Hz) or as a function of position (P) from the casting cooled bottom surface. When the molten alloys reached the *liquidus* temperatures + 5% at the thermocouple closest to the casting's bottom, water flow was activated against the external surface of the mold bottom part, promoting upward directional solidification. Specific information about this experimental setup can be found in previous studies [14, 15].

Various samples obtained under different processing conditions (i.e., different P) were extracted from the DS castings. These samples were appropriately sanded, polished, and etched to expose the resultant microstructures for further examination. After acquiring images using an optical microscope (Olympus Corporation, GX41 model, Tokyo, Japan), the triangle and intercept techniques were used to measure the primary, λ_1 , and secondary, λ_2 , dendritic arm spacings, respectively [30]. Furthermore, a SEM (Philips XL 30 FEG, Eindhoven, The Netherlands) was used to get greater magnification images as well as to identify the produced phases.

Transverse samples associated with a cooling rate of approximately 0.7°C/s were extracted to be used in the LSR experiments. The surface treatment experiments of the four alloys of interest were conducted with an Aurora Labs® S-Titanium Pro machine equipped with a CO_2 laser, which has a maximum power of 300W. The focused laser beam had a nominal spot size of $150\text{ }\mu\text{m}$ with a wavelength of $10.6\text{ }\mu\text{m}$. The laser head is connected to a CNC-controlled XYZ table on which the sample is placed and moved during the treatment. During the LSR trials, the following operating conditions pairs (Q: laser power/v: laser beam speed) were used: 250W/1mm/s, 250W/5mm/s, 250W/10mm/s, 250W/15mm/s, 250W/20mm/s, 300W/1mm/s, 300W/2mm/s, 300W/5mm/s, 300W/8mm/s, 300W/10mm/s, 300W/15mm/s, and 300W/20mm/s. Area energy densities (ED) ranged from 83 to $2,000\text{ J/mm}^2$. Eight EDs were chosen to be evaluated here to alter significantly the examination of laser conditions: 83, 100, 133, 200, 250, 333, 400, and $2,000\text{ J/mm}^2$.

The sample surface was sandblasted before laser treatments to increase energy absorption from the laser beam. Conventional metallographic techniques were used to produce selected transverse (perpendicular to the growth direction) slices of the laser track. Optical (Olympus Corporation, GX41 model, Tokyo, Japan) and SEM (Philips XL30 FEG, Eindhoven, The Netherlands) microscopes were used to examine these sections. Moreover, microstructural spacings were determined within the laser molten pool. Vickers microhardness tests were performed on the cross-sections of the samples using test loads of 25 gf and 500 gf and a dwell time of 15 s.

The sketch in Fig. 1 displays all of the experimental techniques utilized and the details underlying their respective microstructural analyses.

3. Results And Discussion

3.1. Directional solidification: cooling path, microstructures, and scaling laws

The solidification precipitation sequences of the alloys may be described using the CALPHAD method, as shown in Fig. 2. For the AlSi10Mg , AlSi10Mg-1Ni , and AlSi10Mg-2Ni alloys, the $\alpha\text{-Al}$ phase began to solidify first from the liquid. The Al_3Ni phase was the initial phase in the AlSi10Mg-3Ni alloy. The Si, Al_3Ni , and Mg_2Si phases were predicted to precipitate at lower temperatures. The impacts of Ni content change might be noticeable if the fractions of Al_3Ni at 100°C were examined, which were 2.4%, 4.8%, and 7.1% with increasing Ni content. These diagrams in Fig. 2 were also helpful to confirm the experimentally determined *liquidus* and $\alpha\text{-Al}$ precipitation temperatures of 593.8°C , 587.5°C , 583.5°C , and 580.0°C for the AlSi10Mg , AlSi10Mg-1Ni , AlSi10Mg-2Ni , and AlSi10Mg-3Ni alloys, respectively. The computed Si and Mg_2Si fractions did not change substantially with Ni additions.

Optical longitudinal and cross-section images are captured at key spots throughout the length of each casting, based on the evolution of the DS samples, as shown in Fig. 3 and Fig. 4 for the examined alloys. The generation of well-defined dendritic arrays with noticeable changes in length scale as a function of the cooling rate, T , and α -Al growth rate, V , encountered in different casting regions can be seen in Fig. 3 and Fig. 4. The cooling rate range varying between 14 and 0.02 °C/s may be related to changes in both microstructural morphology and length scale [14]. An α -Al matrix with a dendritic pattern (lighter areas in Figs. 3 and 4) and well-defined primary and secondary arms form the microstructure of all examined alloys, with interdendritic areas composed of the: α -Al, Si particles, Mg_2Si , and Al_3Ni intermetallics.

The compositions of phases of the samples solidified at 0.7°C/s were determined by SEM/EDS. These results can be seen at the bottom part of Fig. 5. All phases predicted by the CALPHAD computations were confirmed here. While the Mg_2Si intermetallics has a Chinese-script-like form in the ternary Al-Mg-Si alloy [31], this does not appear to be the case with Ni-containing alloys, where some Mg_2Si particles have a more refined structure, and others have a Chinese-script shape. Moreover, Si particles display a lamellar shape in all evaluated alloys, which is a typical morphology reported for the Al-Si-based alloys solidified under a slow cooling regime [32, 33]. The addition of Ni propitiates the growth of the complex-shaped white Al_3Ni particles, as shown in Fig. 5. Chen and Thomson [34, 35] identified the growth of AlNi-bearing intermetallics in Al-Si-based multicomponent alloys (major additions of Cu, Ni, and Mg) used for pistons, and Mrówka-Nowotnik described these intermetallics to have a complex rod-like shape [36]. Canté et al. [37] observed regular rod-like Al_3Ni in their experiments with hypoeutectic Al-Ni alloys, but the morphology seen here is more complex. Moreover, Jain and Gupta [38] reported that the solubility of Si in the Al_3Ni phase may vary from 0.75–1.66 at.%, which agrees with the present EDS findings.

Jain and Gupta [38] also reported the presence of a ternary eutectic structure in Al-Si-Ni alloys. Acicular Si in equilibrium with Al and bright Al_3Ni in the shape of a skeleton structure has been reported. The morphology of Al_3Ni shown by Jain and Gupta [38] is much closer to that of the current microstructures in Fig. 5.

The thermal profiles obtained through several thermocouples for all examined alloys were analyzed to establish proper plots of position in the casting against the time of the α -Al isotherm passage during solidification: $P = f(t)$. The α -Al growth rate (V_a) was calculated using the time derivatives of these functions, as shown in Fig. 6. In addition, the time derivative of each cooling curve (dT/dt) was computed immediately after the α -Al isotherm has passed through each thermocouple position to estimate the α -Al cooling rate (T) as a function of casting position, as also shown in Fig. 6. When using the transformation isotherm related to the α -Al phase as representative of a larger mass fraction (Fig. 2) for all alloys, it is clear that a wide range of cooling rates exists, ranging from approximately 14 °C/s near the chilled surface to less than 0.1 °C/s near the top of the DS castings.

Considering the very first solidification times (i.e., $P < 5$ mm), it appears that Ni additions have not been able to change the cooling rate. This might be due to small variations in alloy thermophysical properties produced by Ni addition, notably in the thermal and volumetric contraction occurring during solidification. These small variations do not appear to be sufficient to change the alloy/mold heat transfer coefficient (h). A change in h may result in a change in the initial cooling rate profile [39], which is not the case here. The determination of h , on the other hand, is outside the scope of this research.

Figure 7 shows the λ_1 and λ_2 values as a function of cooling rate and growth rate during DS. A single λ_1 -dendritic scaling relation may be typified by a power function for the Ni-containing alloys. The -0.5 exponent described the dendritic development for these alloys in Fig. 7(a), being typical of those reported for multicomponent Al-based alloys [40]. However, a -0.4 exponent has characterized the experimental λ_1 variation for the ternary AlSi10Mg alloy.

Figure 7 (b) shows experimental growth laws relating the λ_2 to the growth rate. This plot has been figured out from the experimental λ_2 results measured along the length of the DS alloy castings. Power functions with an exponent of $-2/3$ characterize the evolution of λ_2 . This exponent is also consistent with those found in λ_2 growth during transient DS of Al-based alloys [29, 41]. The alloy without Ni resulted in a profile of λ_2 higher than that of the other alloys. Due to solute rejection ahead the solid/liquid interface (S/L), the liquid in contact with it has become more solute-enriched. Because Ni solute has no solubility in the α -Al phase, there will be a local composition that differs from the concentration of the available liquid. The establishment of a temperature gradient in the liquid will be caused by this gradient in solute concentration, and higher constitutional supercooling levels will be observed in the areas adjacent to the S/L interface. This might account for the decrease in λ_2 caused by Ni additions.

3.2. LSR: gradient of microstructures and growth mechanisms in the molten pool

The microstructural features of the laser-treated zone and substrate (previously directionally solidified in a water-cooled mold) are depicted in Fig. 8 and Fig. 9. The substrate had a coarse dendritic microstructure, and the resolidified zone (either low or high ED) was characterized by a very fine dendritic arrangement at the molten pool bottom. Furthermore, optical images in Fig. 8 and Fig. 9 depict the shift in microstructure size at the top surface of the molten pool, as indicated by dashed white lines. The layer closest to the base is the bottom layer, while the one above the indicated line is the top layer. In sum, a morphological transition was detected for all examined alloys and conditions. Han and Jiao [42] observed a refinement in the microstructure after post-processing with LSR of the Selective Laser Melted (SLM) fabricated AlSi10Mg alloy surface. Microhardness was observed to be enhanced due to the LSR. These authors reported the formation of a dendritic structure finer than that related to the previous SLM as-fabricated samples.

Wang et al. [43] affirmed that the dendrites might fragment due to Marangoni convection effects giving origin to equiaxed α -Al grains. These fragments may be transported towards the top surface, forming the final refined layer. Theoretical and experimental solidification investigations show that an increase in cooling rate resulted in a change of the micromorphology of the solid/liquid interface from planar to cellular and subsequently to dendritic, implying a direct effect of solidification kinetics on the morphology of the solidification interface. Some authors demonstrated that the dendritic front may be

transformed into a cellular and planar front at higher cooling rates, defining a reverse transition that may produce high-cooling rate cells towards the top [30, 44, 45]. Similar findings have been demonstrated in the literature when SLM has been used to process Al alloys [46, 47].

It is worth noting the formation of a gradient of microstructures towards the top surface of the molten pool in the present investigated samples. Liu et al. [9] estimated top surface and bottom cooling rates of approximately 10^6 °C/s and 10^3 °C/s, respectively, during SLM of the AlSi10Mg alloy at a very high laser speed of 1500 mm/s. These authors stated that studying the gradient of microstructures is essential for laser-treated surfaces since the growth of short-range distinct microstructures may substantially influence on the resulting mechanical properties. The fundamental reason for the gradient of microstructures in terms of Si particle distribution and dendritic/cell size is the variation in molten pool cooling rate [9]. Moreover, the gradient of microstructures with increasing grain size from the surface to the inner has improved mechanical behavior [16].

Analyzing the SEM images of some of the tested conditions in Fig. 8 and Fig. 9, it is observed that lower ED refers to the growth of rounded cells or grains, while more complex structures (dendrites) characterize the top surface of high-ED samples.

It is feasible to evaluate the distribution of elements after laser processing using the color mapping data for the AlSi10Mg-3Ni alloy surface solidified under low ED conditions. Mg and Fe appear to be equally distributed throughout the microstructure, but Ni and Si remain concentrated at the grain boundaries. It seems that the present solidification conditions allowed rejection of solute to happen in such a way that second phases such as Al_3Ni and Si could be formed at the end stage of solidification.

The greater tendency of Ni and Si to segregate in the solidification front may be linked to the lower redistribution coefficients (k) related to these elements, 0.008 [37] and 0.12 [48], respectively. On the other hand, k for the Al-Mg system is approximately 0.5 [49], which may inhibit its accumulation at the solidification interface. In addition, Mg is less dense than Al, and because of this, it can be better distributed in the remaining liquid during rapid solidification. Furthermore, the impurity Fe level is around 0.1 wt%, favoring an even distribution of this element, as seen in Fig. 10.

The primary spacing varied from an average value of approximately 440 μm in the non-treated zone to a mean spacing of approximately 1.8 μm in the center part of the resolidified zones of the investigated alloys, i.e., a remarkable decrease in length scale. The equation proposed by Ashby and Easterling [50] allowed for the cooling rate in laser melting traces to be estimated, based on the following expression:

$$\dot{T} = \frac{-2\pi k (T_a - T_0)^2}{A \left(\frac{Q}{v} \right)}, \quad (1),$$

where A is the laser beam absorptivity for the laser beam (0.18 as stated by Liu et al. [51]), v (m/s) is the laser scanning speed, Q (W) is the laser power, k (W/mK) is the liquid thermal conductivity determined through CALPHAD, T_0 is the room temperature, and T_a is the transformation temperature of the α -Al growth start.

Given that the thermal conductivity and T_a changed minimally as a function of Ni addition, the calculated cooling rates were similar for the alloys studied. For instance, $\dot{T}$ was about 1.3×10^4 °C/s for the 300W/5mm/s and 5.0×10^4 °C/s for the 300W/20mm/s. Extrapolations from the scaling law of average λ values of 2.47 μm and 1.38 μm for the 3 Ni-containing alloys under the same aforementioned conditions resulted in estimated molten pool cooling rates of 1.5×10^4 °C/s and 4.7×10^4 °C/s, which were adequately validated for the 300W/5mm/s and 300W/20mm/s LSR conditions, respectively. The extrapolation approach, which permits estimating the point-to-point cooling rate inside the molten pool, was validated by the closeness of the cooling rate values determined by both methods.

Table 1 shows data for the 24 conditions, 6 for each alloy. The dimensions of the molten pool as well the fractions of the bottom and top layers may be observed. Average hardness values are also included. Although the spacing values at the bottom and center varied relatively little, an important impact on the extrapolated cooling rates is attained if those regions are compared. The extrapolation was performed considering information related to the bottom and center of the molten pool since the top layer is very affected by convection, and the λ scaling relationships are not valid for this layer.

Flow dynamics is a feature of molten pools that is sometimes neglected. The flow in a molten pool is primarily caused by a spatial fluctuation in surface tension, called thermocapillary flow or Marangoni convection. Fluid flow has been demonstrated to have a crucial role in heat transfer and solidification behavior in the molten pool, affecting solidification microstructure, alloy element distribution, and defects in previous researches [52–54]. In a diffusive-driven regime, the growth rate is expected to increase continuously toward the pool top, and the effect on dendrite refinement would be even higher, producing lower λ [3]. However, this is not the case if the present results for spacing are considered. Most conditions tested show that the spacing is larger at the bottom, decreasing at the center and increasing again towards the top surface. A mechanism explaining the reversion of λ across the molten pool can be seen in Fig. 11.

It appears that the induced convection in the molten pool affects growth by generating fragmentation (Fig. 11b). In the central part of the pool, despite convection, solidification velocities are still higher than those at the bottom, resulting in lower λ (Fig. 11c). However, due to solute drag from the interdendritic regions to the top and the presence of floating fragments, the λ characterizing of the top surface tends to be increased (Fig. 11d).

Some of the conditions increased λ from the bottom towards the top, especially for higher Ni content and low laser speed. Reduced laser speed causes a significant increase in contact duration and peak temperature, resulting in greater Marangoni fluid convection. These features appear to explain the

increasing λ profile in a few cases. Indeed, low laser speed conditions seem to provide a more significant accumulation of solute in the upper part, resulting in a thicker eutectic film, according to Fig. 12.

Table 1
Data extracted for all examined laser treated AlSi10Mg(-Ni) alloy surfaces.

Alloy	Q [W]	v [mm/s]	Molten pool size (width x depth) [μm]	Energy density [J/mm ²]	Bottom layer fraction [%]	Top layer fraction [%]	Microstructure Spacing [μm]			Cooling rate at the bottom [K/s]	Cooling rate before top layer [K/s]	Theoretical cooling rate [K/s]	Molten pool average Hardness [HV]
							λ_b	λ_c	λ_t				
							[bottom]	[center]	[top]				
Non-modified	300	1	463 x 155	2000	77.7	22.3	2.1	2.6	1.4	5.5x10 ⁴	3.2x10 ⁴	2.5x10 ³	100.8
	300	5	638 x 264	400	73.8	26.2	2.8	2.4	2.6	2.7x10 ⁴	4.0x10 ⁴	1.3x10 ⁴	105.9
	300	8	633 x 267	250	71.4	28.6	2.5	1.8	2.3	3.5x10 ⁴	7.5x10 ⁴	2.0x10 ⁴	107.8
	300	10	651 x 274	200	73.3	26.7	2.2	1.9	2.5	4.9x10 ⁴	7.0x10 ⁴	2.5x10 ⁴	110.7
	300	15	632 x 257	133	73.2	26.8	2.2	1.7	2.0	4.9x10 ⁴	8.8x10 ⁴	3.8x10 ⁴	113.2
	300	20	621 x 240	100	54.8	45.2	2.1	1.5	2.0	5.4x10 ⁴	1.2x10 ⁵	5.0x10 ⁴	113.8
Modified with 1%Ni	250	5	472 x 180	333	68.4	31.6	2.4	1.9	2.6	1.6x10 ⁴	2.4x10 ⁴	1.5x10 ⁴	122.2
	250	20	446 x 157	83	84.4	15.6	1.4	1.0	1.3	4.3x10 ⁴	8.8x10 ⁴	6.1x10 ⁴	126.7
	300	1	475 x 175	2000	89.0	11.0	3.2	2.9	2.3	8.8x10 ³	1.1x10 ⁴	2.5x10 ³	121.6
	300	5	550 x 216	400	72.3	27.7	2.5	2.0	2.7	1.4x10 ⁴	2.3x10 ⁴	1.3x10 ⁴	121.7
	300	10	520 x 205	200	79.9	20.1	1.6	1.4	2.0	3.4x10 ⁴	4.7x10 ⁴	2.5x10 ⁴	124.3
	300	20	508 x 190	100	66.5	33.5	1.6	1.4	1.5	3.6x10 ⁴	4.9x10 ⁴	5.0x10 ⁴	127.5
Modified with 2%Ni	250	5	467 x 179	333	64.2	35.8	2.1	1.5	2.2	2.0x10 ⁴	4.1x10 ⁴	1.5x10 ⁴	130.9
	250	20	408 x 140	83	85.2	14.8	1.7	1.1	1.4	3.2x10 ⁴	8.5x10 ⁴	6.0x10 ⁴	127.0
	300	1	492 x 177	2000	84.8	15.2	2.9	3.2	3.4	1.1x10 ⁴	8.9x10 ³	2.5x10 ³	131.7
	300	5	521 x 193	400	83.1	16.9	1.9	1.9	2.4	2.6x10 ⁴	2.4x10 ⁴	1.3x10 ⁴	126.2
	300	10	512 x 182	200	83.3	16.7	1.9	1.5	2.2	2.6x10 ⁴	4.2x10 ⁴	2.5x10 ⁴	131.3
	300	20	501 x 182	100	87.9	12.1	1.8	1.1	1.3	2.7x10 ⁴	8.2x10 ⁴	5.0x10 ⁴	142.6
Modified with 3%Ni	250	5	505 x 195	333	71.7	28.3	2.1	1.6	2.1	2.1x10 ⁴	3.6x10 ⁴	1.5x10 ⁴	129.9
	250	20	429 x 173	83	67.1	32.9	1.7	1.2	1.4	3.0x10 ⁴	6.2x10 ⁴	6.0x10 ⁴	132.2
	300	1	533 x 201	2000	83.9	16.1	2.7	2.9	3.9	1.2x10 ⁴	1.1x10 ⁴	2.5x10 ³	126.2
	300	5	595 x 231	400	82.9	17.1	2.1	1.9	2.3	2.0x10 ⁴	2.4x10 ⁴	1.3x10 ⁴	133.8
	300	10	553 x 219	200	78.2	21.8	1.8	1.4	1.9	2.7x10 ⁴	4.3x10 ⁴	2.5x10 ⁴	136.7
	300	20	513 x 177	100	75.6	24.4	1.5	1.0	1.3	4.0x10 ⁴	8.5x10 ⁴	5.0x10 ⁴	136.1

When the AlSi10Mg(-Ni) alloys were produced at an energy density range of 83 to 133 J/mm², the hardness resulted is higher, attaining approximately 140 HV for the alloys having higher Ni content, as can be seen in Fig. 13. It appears that as the Ni content is increased, the hardness tends to become constant.

For each condition, hardness measurements were carried out at the bottom, center, and top of the molten pool, generating an appreciable set of hardness as a function of λ for each alloy, as shown in Fig. 14. Although λ has no effect on the hardness for all examined alloys, the LSR treatment has a notable influence on the hardness of all treated surfaces of the Al-Si-Mg-Ni samples. The hardness of the remelted zone increased from 75–80 HV in the substrate to 107–131.5 HV. The fineness of the eutectic mixture and the resulting homogenous distribution of intermetallic particles, which resulted from the laser treatment, are primarily responsible for the greater hardness values of the remelted zones.

4. Conclusions

The microstructures of the DS samples of any alloy examined have generated well-defined α -Al dendritic arrays with noticeable changes in length scale as a function of cooling rate, T , and α -Al growth rate, V along the length of the castings. The dendritic pattern was shown to be characterized by well-defined primary (λ_1) and secondary (λ_2) arms with interdendritic areas composed of: α -Al; Si particles; Mg₂Si, and Al₃Ni intermetallics. Scaling power function laws relating λ_1 to T and λ_2 to V were experimentally determined for any alloy examined. The alloy without Ni resulted in a profile of λ_2 higher than those of the other alloys

The laser-treated zone was shown to be formed by a gradient of microstructures towards the top surface of the molten pool, characterized by a very fine dendritic arrangement at the bottom/center of the molten pool, followed by a shift in microstructure size at the top surface for all examined alloys. The solidification conditions allowed rejection of solute to happen in such a way that second phases such as Al₃Ni and Si could be formed at the end stage of solidification. λ_1 varied from an average value of approximately 440 μ m in the non-treated zone to a mean spacing of approximately 1.8 μ m in the center part of the resolidified zones.

The laser surface remelting treatment (LSR) was shown to have a remarkable influence on hardness of all treated surfaces of the Al-Si-Mg-Ni samples. The hardness of the remelted zone increased from 75–80HV at the substrate to 107–131.5HV. The fineness of the eutectic mixture and the resulting homogenous distribution of intermetallic particles, which resulted from the laser treatment, are the primarily responsible for the greater hardness of the remelted zones.

Declarations

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Availability of data and material

Data presented in this study are available on request from the corresponding author. Data are not publicly available because they pertain to a research still in development.

Ethical approval The authors declare that there is no ethical issue applied to this article.

Consent to participate The authors declare that all authors have read and approved to submit this manuscript to IJAMT.

Competing interests The authors declare that there is no conflict of interest.

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Figures

Figure 1

Operational sequence of production of directionally solidified castings, characterizations, and laser surface treatment.

Figure 2

Equilibrium profiles for the a) AlSi10Mg, b) AlSi10Mg-1Ni, c) AlSi10Mg-2Ni, and (d) AlSi10Mg-3Ni alloys through CALPHAD calculations.

Figure 3

Longitudinal optical microstructures for the DS AlSi10Mg-(Ni) alloys.

Figure 4

Transversal optical microstructures for the DS AlSi10Mg-(Ni) alloys.

Figure 5

SEM typical microstructures and EDS results for the DS AlSi10Mg-(Ni) alloys.

Figure 6

Variations of α -Al (a) cooling rate, and (b) growth rate from the lower to the upper regions of the AlSi10Mg-(Ni) alloys castings.

Figure 7

(a) Primary and (b) secondary dendrite arm spacing as a function of cooling rate and growth rate. R^2 is the coefficient of determination.

Figure 8

Cross-section microstructures of the laser-treated (a) AlSi10Mg, (b) AlSi10Mg-1Ni, (c) AlSi10Mg-2Ni, and (d) AlSi10Mg-3Ni alloy surfaces for an ED of approximately 400 J/mm².

Figure 9

Cross-section microstructures of the laser-treated (a) AlSi10Mg, (b) AlSi10Mg-1Ni, (c) AlSi10Mg-2Ni, and (d) AlSi10Mg-3Ni alloy surfaces for an ED of approximately 100 J/mm².

Figure 10

EDS elemental mapping images display elements' distribution in the remelted AlSi10Mg-3Ni alloy surface for an ED of approximately 100 J/mm².

Figure 11

Scheme showing the microstructure evolution and flow dynamics effects across the molten pool explaining I reversion towards the top: (a) completely molten pool; (b) initial growth of dendrites towards the top; (c) growth of finer dendrites in the central area and transportation of solid fragments; and (d) end-stage of solidification forming the top layer. The subscripts b, c and t in the illustrative drawing mean bottom, center and top of the molten pool.

Figure 12

Example of laser remelting AlSi10Mg-3Ni alloy surface resulting in coarse eutectic structure on the top surface: 300W /1 mm/s; ED of 2000 J/mm².

Figure 13

Vickers hardness variation as a function of Ni content and ED for the laser-treated AlSi10Mg(-Ni) alloys.

Figure 14

Molten pool hardness as a function of the microstructural spacing for the laser remelted AlSi10Mg(-Ni) alloy surfaces: (a) AlSi10Mg, (b) AlSi10Mg-1Ni, (c) AlSi10Mg-2Ni, and (d) AlSi10Mg-3Ni.