

Article

Magnesium Solubility in Primary α -Al and Heat Treatment Response of Cast Al-7Si-Mg

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Abstract: Magnesium and silicon concentrations in the interior of primary α -Al of Al-7Si-Mg alloys were studied at temperatures in the liquid-solid range and just after solidification was completed. Analysis of the results showed that the magnesium concentration in the interior of primary α -Al is very low in the temperatures range from the liquidus to the start of the Al-Si eutectic reaction. Formation of silicon-rich phases during eutectic reactions, such as eutectic silicon and β -Al₅FeSi, phases trigger a significant increase in the magnesium concentration in the interior of primary α -Al, when sufficient time is allowed for solid-state diffusion to occur. When fast cooling rates are applied during the Al-Si eutectic reaction, most of the magnesium is retained in π -Al₈FeMg₃Si₆ and Mg₂Si phases formed during solidification. Semi-solid Al-7Si-Mg castings were produced with varying magnesium contents, and the mechanical properties were evaluated in the as-cast, T5 and T6 conditions. It was found that the 0.2% offset yield strength of the semi-solid Al-7Si-Mg castings in the T5 and T6 conditions increases linearly with the square root of the magnesium concentration in the interior of the α -Al globules formed during the slurry preparation process.

Keywords: aluminium alloys; solid-solution; heat treatment; solubility; rheocasting; mechanical properties

1. Introduction

Al-7Si-Mg alloys are commonly used in the automotive and aerospace industries due to their excellent castability, high specific strength and corrosion resistance [1–4]. The addition of up to 0.5 wt.% magnesium to these alloys improves the response to precipitation hardening during aging of these alloys [2,5–7]. However, magnesium addition decreases ductility and fracture toughness [8]. Typically, Al-7Si-Mg castings have relatively low strength in the as-cast condition and, for this reason, castings produced using alloys of the series 356 and 357, i.e., the two most used Al-7Si-Mg alloys, are commonly heat treated to the T6 condition to achieve the desired mechanical properties [1,9].

The effectiveness of T5 and T6 heat treatments of Al-7Si-Mg alloys is strongly affected by the level of silicon and magnesium supersaturation of the primary α -Al at the artificial aging temperature [1]. Increasing the supersaturation of magnesium and silicon in α -Al will promote the formation of larger number density and a fraction of metastable and coherent β'' precipitates, and enhance the hardening response of the Al-7Si-Mg alloys during aging [1]. In the T6 heat treatment, a solution treatment is carried out to homogenize the primary α -Al, dissolve intermetallic phases such as Mg₂Si and π -Al₈FeMg₃Si₆, releasing magnesium into solid solution, and spheroidize eutectic silicon [2,3].

In the T5 heat treatment, the as-solidified casting is artificially aged, and a moderate increase of strength is obtained without the risk of distortion [5]. Additionally, the T5 heat treatment is less expensive and blistering is prevented compared to T6 heat treatment [5,6]. However, the strength of castings in the T5 condition is generally significantly lower, compared to castings in the T6

condition [9,10]. In Al-7Si-Mg castings, the increase of magnesium content of the alloy results in an increase of the strength achieved after T5 or T6 heat treatments [7]. However, it is critical that the cooling rate is high in the range of temperatures 400 to 260 °C to avoid premature precipitation that is detrimental to mechanical properties [8].

The cooling rate during solidification and subsequent solid-state cooling influence solute concentrations and distributions in the interior of α -Al dendrites [11,12]. Sjölander and Seifeddine [13] measured solute concentrations in the interior of α -Al dendrites in directionally solidified Al-7Si-0.3Mg samples that produced different Secondary Dendrite Arm Spacings (SDAS). A total of 0.04 wt.% magnesium and 1.5 wt.% silicon were measured in the interior of α -Al dendrites of an Al-7Si-0.3Mg casting with SDAS of 10 μm . However, for a casting with SDAS of 51 μm , i.e., lower cooling rate, the magnesium concentration in the interior of α -Al dendrites increased to 0.1 wt.%, while the silicon concentration decreased to 1.4 wt.%, in comparison to the castings produced with a SDAS of 10 μm . Pedersen and Arnberg [11] measured a concentration of 0.05 wt.% magnesium and 1.1 wt.% silicon in the interior of the α -Al secondary dendrite arms of an as-cast Al-7Si-0.24Mg casting with SDAS of 13 μm . For the same alloy cast under solidification conditions that produced a SDAS of 52 μm , the magnesium and silicon concentrations in the interior of the α -Al secondary dendrite arms were 0.1 wt.% and 1.5 wt.%, respectively. Rometsch et al. [14] measured the magnesium and silicon concentrations in the interior of α -Al dendrites of sand-cast Al-7Si-0.4Mg with SDAS of 40 μm and obtained 0.13 wt.% and 1.37 wt.%, respectively. For sand cast Al-7Si-0.62Mg with a SDAS of 55 μm , the magnesium and silicon concentrations in the interior of the α -Al dendrites were 0.25 wt.% and 1.63 wt.%, respectively. Therefore, the magnesium concentration in the interior of α -Al dendrites of as-cast Al-7Si-Mg alloys decreases with the increase of cooling rate, i.e., smaller SDAS and increases with the increase of the magnesium content of the alloy. After solution treatment, the magnesium concentration in the interior of primary α -Al dendrites increases markedly compared to that obtained in as-solidified Al-7Si-0.3Mg castings with varying SDAS [13]. However, the silicon concentration in the interior of primary α -Al dendrites after solution treatment can be even lower, compared to the same Al-7Si-0.3Mg casting in the as-cast condition [13].

Most of the studies of heat treatment response of Al-7Si-Mg castings are focused on the dissolution of intermetallics, spheroidization of eutectic silicon and homogenization of primary α -Al during solution treatment of varying as-cast microstructures [13,15–17]. The solution treatment efficacy will determine the subsequent precipitation hardening response during aging and mechanical properties achieved of a specific T6 heat treated Al-7Si-Mg alloy. These studies are carried out in the solid-state. However, very little information is found about the dissolution of solute elements in the interior of α -Al in the liquid-solid range, which is of particular importance for T5 heat treatment response of Al-7Si-Mg castings. In this study, the solubility of magnesium and silicon in the primary α -Al was investigated at temperatures within the solidification range and just after the end of the solidification of Al-7Si-Mg alloys. The results are compared to thermodynamic calculations. Additionally, the magnesium and silicon concentrations in the interior of the α -Al globules of Semi-Solid Metal (SSM) Al-7Si-Mg castings in the T5 and T6 conditions were studied. The tensile properties of SSM Al-7Si-Mg castings were evaluated and correlated to the magnesium concentration in the interior of the α -Al globules formed during slurry preparation.

2. Materials and Methods

2.1. Commercial Al-7Si-Mg Alloys

Commercial Al-7Si-Mg ingots with varying magnesium contents were melted at 700 °C in an electrical resistance furnace. The chemical composition range of the commercial alloys after melting is shown in Table 1. At least two batches of each commercial alloy were used to produce SSM castings and conduct isothermal holding treatments at temperatures above the start of the Al-Si eutectic reaction.

2.2. Unmodified Al-7Si-Mg Alloys

The unmodified alloys were produced by introducing the intended amounts of aluminum, silicon and magnesium in a graphite-bonded silicon carbide crucible and melted at 750 °C, using an electrical resistance furnace. After melting, each alloy was held at 750 °C for 30 min for homogenization. The average chemical composition of the unmodified Al-7Si-Mg alloys is shown in Table 1. These unmodified Al-7Si-Mg alloys were produced to carry out isothermal holding treatments at temperatures close/within the Al-Si eutectic reaction interval. Strontium additions could hinder the formation of Al-Si eutectic during the isothermal holding treatment and for this reason, unmodified alloys were used.

Table 1. Average chemical composition of the Al-7Si-Mg alloys used in this study. Compositions are in weight %.

Type	Alloys	Si	Mg	Fe	Ti	Sr	Al
Commercial	Al-7Si-0.3Mg	6.94–7.10	0.30–0.35	0.13–0.16	0.072–0.099	0.012–0.027	Bal.
	Al-7Si-0.45Mg	7.24–7.39	0.47–0.50	0.11	0.096–0.11	0.024–0.025	Bal.
	Al-7Si-0.6Mg	6.94–7.28	0.59–0.60	0.11–0.12	0.11–0.12	0.023–0.025	Bal.
Unmodified	Al-7Si-0.3Mg	6.58	0.31	0.20	<0.0004	<0.0001	Bal.
	Al-7Si-0.45Mg	7.02	0.44	0.18	<0.0004	<0.0001	Bal.
	Al-7Si-0.6Mg	7.03	0.58	0.20	<0.0004	<0.0001	Bal.

2.3. Isothermal Holding Treatments

After melting, the commercial and unmodified alloys were cast in pre-heated graphite-bonded silicon carbide crucibles for the isothermal holding treatments. Cylinders of about $\phi 35$ mm diameter and 40 mm height were machined from the cast samples using a lathe. Three isothermal holding treatments were carried out at 605 °C, 590 °C and 580 °C for 120 h on cylinders of each commercial alloy. The unmodified Al-7Si-Mg cylinders were isothermally held for 120 h at temperatures close/within the Al-Si eutectic reaction interval, 575 °C, 570 °C and 560 °C. For a selected isothermal holding treatment temperature, one cylinder of each commercial/unmodified Al-7Si-Mg alloy was introduced into a graphite bonded silicon carbide crucible and placed inside an electrical resistance furnace, as shown in Figure 1. The furnace was heated at a rate of about 10 °C/min to 700 °C, held for 15 min, and then allowed to cool to the selected isothermal holding treatment temperature. When the isothermal holding time was completed, the samples were plunged into a water bath at room temperature. A type N thermocouple was used to measure the temperature, due to its better thermal stability during extended exposure to high temperature compared to type K [18]. The thermocouple, calibrated against pure aluminum, was placed in the interior of the furnace to control the temperature, as shown in Figure 1. The calibration temperature was set to 660 °C. The chemical composition of the alloys was controlled before and after isothermal holding treatments using a Spectro Max CCD LMXM3 spectrometer (SPECTRO Analytical Instruments GmbH, Kleve, Germany) to evaluate possible magnesium fading during treatments, and no significant differences were obtained.

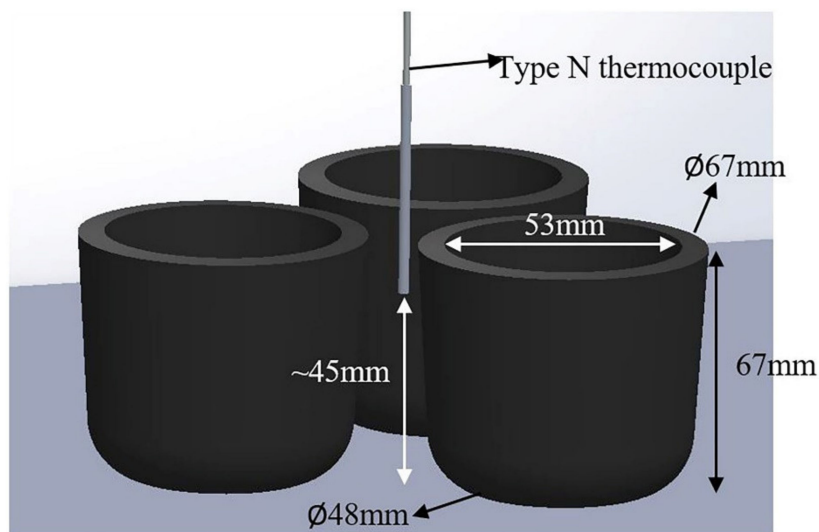


Figure 1. Experimental setup used for the isothermal holding treatments.

2.4. SSM Al-7Si-Mg Castings

The Rheometal™ (Rheometal AB, Stockholm, Sweden) [19] process was used to produce the SSM castings from each commercial alloy shown in Table 1. First, cylinders were cast by pouring the commercial alloys at about 700 °C into a copper mold, with a cavity of $\phi 40$ mm diameter and 100 mm depth. The cast cylinders are termed Enthalpy Exchange Material (EEM) [20]. A 12 mm diameter stainless steel rod was cast about 10 mm deep into the EEMs base. During EEMs production, water circulated through internal channels in the copper mould to maintain reproducible thermal conditions between cast EEMs. After casting, the EEMs height was adjusted to generate an addition of 7 wt.% of the total shot weight. The EEM preheated to 200 °C was immersed while rotating at 850 rpm into about 1.3 kg of commercial alloy at 650 °C (~35 °C superheat), to produce the slurry. The slurry preparation took approximately 18 s. Next, the prepared slurry was immediately poured into the shot-sleeve of a 50 tonne vertical high pressure die casting machine, to produce tensile bars, as shown in Figure 2. A plunger advance speed of 0.3 m/s and an intensification pressure of 160 bar were kept constant in all experiments. A duration of 10 s was set for each stage, intensification pressure and subsequent cooling of the casting in the die-cavity. The SSM castings intended to be aged to the T5 condition were immediately extracted from the die cavity after the intensification pressure stage, and plunged into a water bath at room temperature to minimize precipitation during cooling. The die temperature was controlled by internal oil circulation set at a constant temperature of 175 °C, using a PolyTemp HTF 300 heater (ICS Tricool Thermal, Southampton, UK). A first set of castings were produced, to warm up the assembly and maintain reproducible thermal conditions in the shot sleeve and die-cavity.

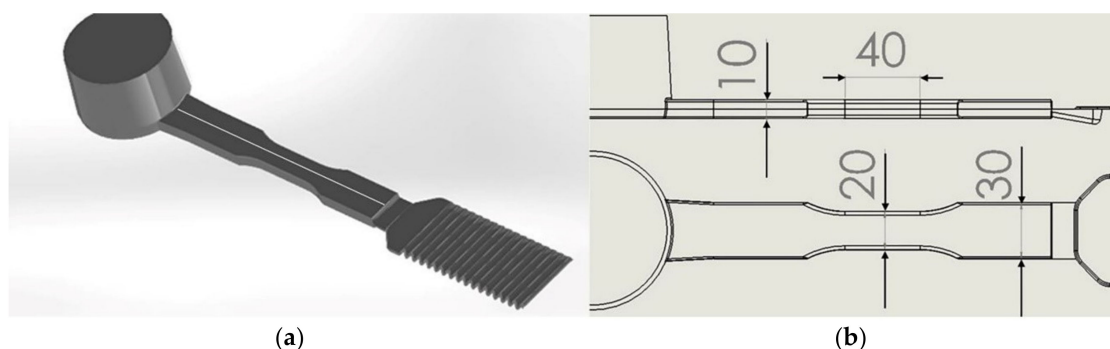


Figure 2. (a) Casting shape; (b) cast tensile bar. Dimensions in mm.

2.5. T6 and T5 Heat Treatments of the SSM Al-7Si-Mg Castings

After casting, some of the cast tensile bars were heat treated to the T5 condition, and others to the T6 condition. For the T5 heat treatment, the SSM Al-7Si-Mg cast tensile bars were artificially aged within a period of 24 h after casting. The T6 heat treatment consisted of a solution treatment, followed by quenching in water at room temperature and artificial aging. The temperatures and holding times used for both T5 and T6 heat treatments for all SSM Al-7Si-Mg cast tensile bars are shown in Table 2. The holding times shown in Table 2 are the times that the SSM Al-7Si-Mg tensile bars were held at the solution treatment and artificial aging temperatures. For both solution treatment and artificial aging, the heating rate of the tensile bars was about 10 °C/min.

Table 2. Temperatures and isothermal holding times used for Solution Treatment (ST) and Artificial Aging (AA) of the different SSM Al-7Si-Mg castings.

Casting	Condition	Temperatures and Holding Times
SSM Al-7Si-0.3Mg	T5	AA at 175 °C for 4.5 h
	T6	ST at 510 °C for 4 h + AA at 190 °C for 2 h
SSM Al-7Si-0.45Mg	T5	AA at 175 °C for 4.5 h
	T6	ST at 510 °C for 4 h + AA 190 °C for 2 h
SSM Al-7Si-0.6Mg	T5	AA at 180 °C for 4.5 h

2.6. Microstructural Characterization

Longitudinal cross-sections of the SSM Al-7Si-Mg castings and from the isothermal holding treatment samples were ground, and the last step of polishing was completed with OP-U suspension before observation. An Olympus GX71F optical microscope (Olympus Corporation, Tokyo, Japan) and a JEOL 7001F (JEOL Ltd., Tokyo, Japan) Scanning Electron Microscope (SEM), equipped with Wavelength-Dispersive Spectroscopy (WDS), were used for microscopy, and to measure silicon and magnesium concentrations in the interior of primary α -Al, respectively. An area of 225 μm^2 in the interior of five different primary α -Al crystals for each casting was measured at a fixed acceleration voltage of 10 kV. Pure aluminum, silicon and magnesium were used as standards. For all WDS measurements, the total concentration of aluminum, silicon and magnesium in primary α -Al crystals was $100 \pm 2\%$. The WDS analysis of the SSM casting samples focused on the α -Al globules that likely formed during slurry preparation (diameter > 25 μm) and are denoted α_1 -Al globules in this study. Energy-dispersive X-ray Spectroscopy (EDS) mapping was conducted at a fixed acceleration voltage of 15 kV, to identify the intermetallic phases formed in the different samples.

2.7. Tensile Testing

Tensile testing was performed at room temperature according to the SS-EN ISO 6892-1:2016 [21]. No surface treatment or machining was performed on the specimens. The tests were carried out in a Zwick/Roell Z100 testing machine (ZWICKRoell Group, Ulm-Einsingen, Germany) at a constant strain rate of 0.00025 s^{-1} , up to the offset 0.2% yield strength, where the strain rate was increased to 0.002 s^{-1} , and kept constant until fracture. The elongation was measured continuously by a laser Zwick/Roell LaserXtens extensometer (ZWICKRoell Group, Ulm-Einsingen, Germany). At least three samples were tested for each condition.

3. Results

3.1. Isothermal Holding Treatments

Figure 3 shows representative microstructures of commercial Al-7Si-Mg samples isothermally held for 120 h at temperatures above the Al-Si eutectic reaction temperature, i.e., 605 °C, 595 °C and 580 °C.

Large primary α -Al of near-globular shape surrounded by fine primary α -Al dendrites and eutectic are obtained in all samples. The fine primary α -Al dendrites and eutectic most likely formed during quenching. No significant differences in microstructure can be observed between the different alloys held isothermally at the same temperature. However, increasing the isothermal holding temperature resulted in a decrease in the fraction of primary α -Al globules and an increase of the fraction of fine primary α -Al dendrites formed for all alloys, as seen in Figure 3.

Figure 4 shows representative microstructures of the unmodified Al-7Si-Mg samples isothermally held at temperatures close/within the Al-Si eutectic reaction temperature interval, i.e., 575 °C, 570 °C and 560 °C. The microstructures of the samples isothermally held at 575 °C show coarse primary α -Al globules, surrounded by fine primary α -Al dendrites and eutectic, as shown in Figure 4. A decrease of the isothermal holding temperature from 580 to 575 °C resulted in a less globular morphology of the coarser primary α -Al, particularly for the unmodified Al-7Si-0.3Mg and Al-7Si-0.45Mg alloys, as shown in Figures 3 and 4. Jarfors [22] observed, for an A356 alloy, that the shape factor of the α -Al phase decreased with the decrease of the isothermal holding temperature, particularly for temperatures lower than 585 °C, which correlates well with the observations in this study. Additionally, entrapped eutectic was observed in all samples isothermally held at 575 °C, but was significantly less in samples isothermally held at higher temperatures, as shown in Figures 3 and 4. It seems that the fraction of fine primary α -Al dendrites after isothermal holding treatment at 575 °C was lower, compared to the samples isothermally held at 580 °C.

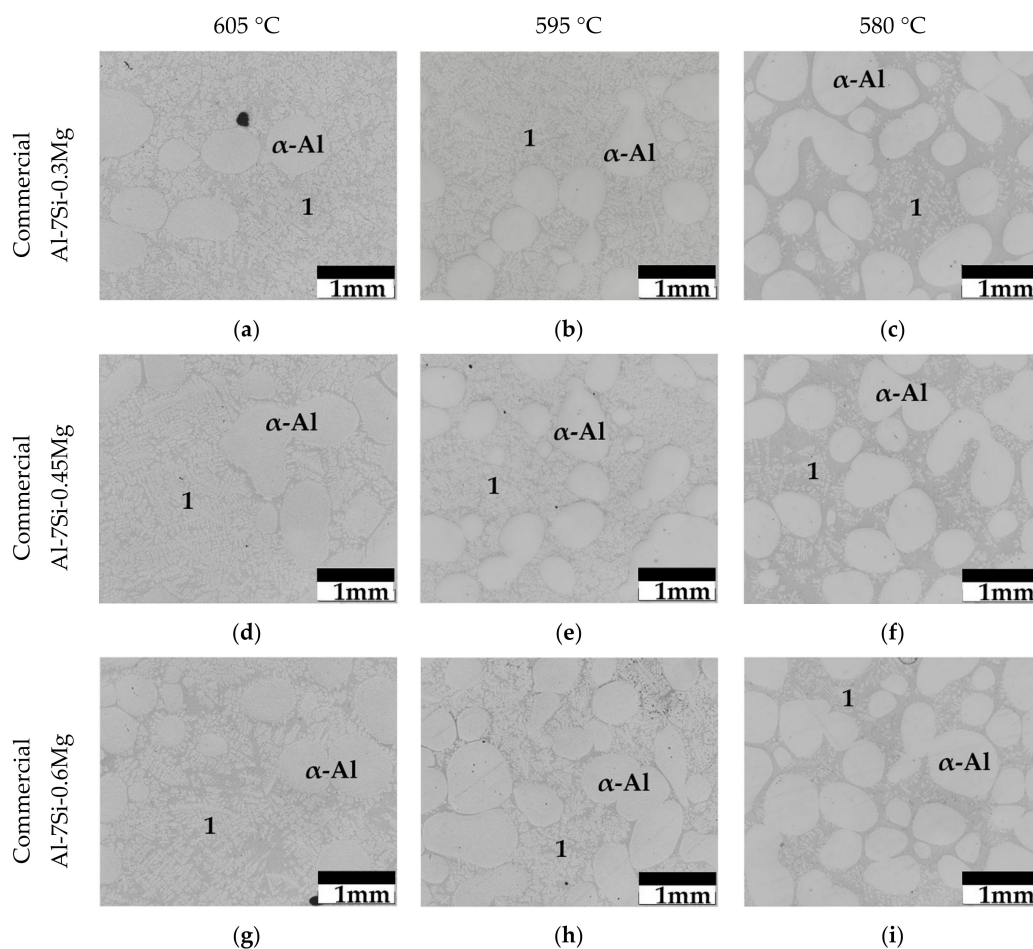


Figure 3. Micrographs of the commercial Al-7Si-Mg alloys quenched after isothermally holding for 120 h at temperatures between the nucleation of the first solid and main Al-Si eutectic reaction: (a–c) Al-7Si-0.3Mg, (d–f) Al-7Si-0.45Mg and (g–i) Al-7Si-0.6Mg. Region 1: primary α -Al dendrites + eutectic.

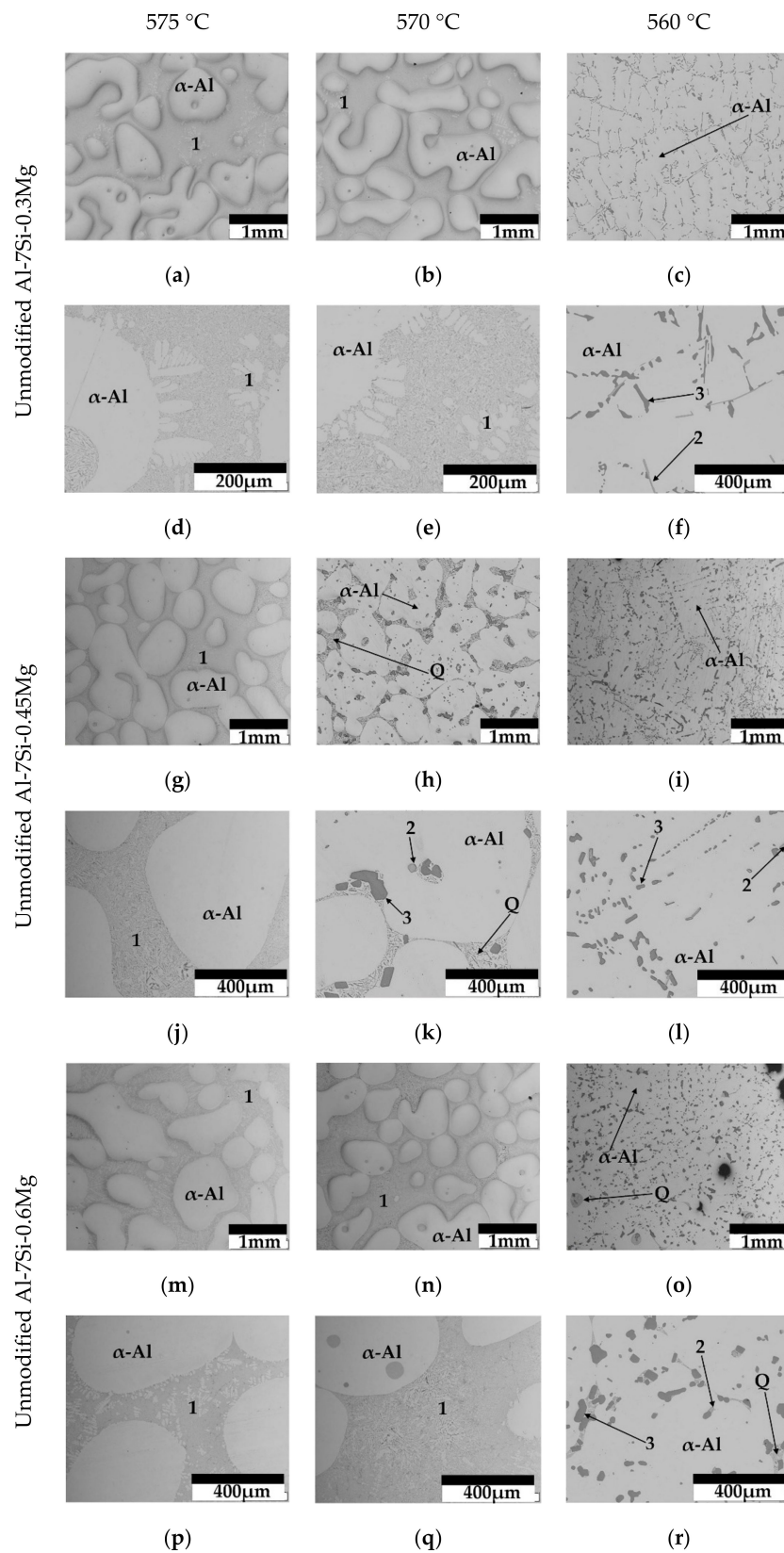


Figure 4. Micrographs of the unmodified Al-7Si-Mg alloys held isothermally at 560 °C, 570 °C and 575 °C during 120 h: (a–f) Al-7Si-0.3Mg, (g–l) Al-7Si-0.45Mg and (m–r) Al-7Si-0.6Mg alloys. Region 1: primary α -Al dendrites + eutectic. Region 2: iron-rich intermetallic. Region 3: eutectic silicon. Q: quenched liquid.

The eutectic microstructure changed significantly with the decrease of the isothermal holding temperature below 575 °C, as shown in Figure 4 for the unmodified Al-7Si-0.45Mg sample isothermally held at 570 °C, and all samples isothermally held at 560 °C. Figure 5 shows silicon, magnesium and iron distributions in the eutectic regions of the samples isothermally held at 570 °C and 560 °C. The intermetallic phases were determined by analysis of EDS mapping and the literature [13,16,23,24]. The π -Al₈FeMg₃Si₆ phase was identified by containing a higher concentration of magnesium, silicon and iron, compared to the surroundings, as shown in Figure 5q–t. The β -Al₅FeSi phase was distinguished from the π -Al₈FeMg₃Si₆ phase by the absence of magnesium, and Mg₂Si was the darkest phase observed in SEM micrographs, as shown in Figure 5q. Similar microstructures were obtained for the unmodified Al-7Si-0.3Mg and Al-7Si-0.6Mg samples isothermally held at 570 °C, compared to samples of the same alloys isothermally held at 575 °C. In the eutectic regions, the unmodified Al-7Si-0.3Mg sample showed very fine Al-Si eutectic, β -Al₅FeSi and π -Al₈FeMg₃Si₆ phases after isothermal holding treatment at 570 °C. The increase of magnesium content of the alloy from 0.3 wt.% to 0.6 wt.% resulted in the formation of π -Al₈FeMg₃Si₆ and Mg₂Si phases in the eutectic regions in addition to fine Al-Si eutectic after isothermal holding treatment at 570 °C, as shown in Figure 5q–t). However, the unmodified Al-7Si-0.45Mg sample isothermally held at 570 °C showed quenched liquid regions, and other regions consisting of a coarse eutectic microstructure that most likely formed during the isothermal holding at 570 °C, as seen in Figure 4h,k. The coarse eutectic regions contained large eutectic silicon and β -Al₅FeSi phases, while the quenched liquid regions consisted of very fine eutectic silicon flakes, π -Al₈FeMg₃Si₆ and Mg₂Si phases, as shown in Figures 4k and 5i–l. Coarsening of eutectic silicon and β -Al₅FeSi phases occurred during isothermal holding at 570 °C, which suggests that the binary $L \rightarrow Al + Si$ and ternary $L \rightarrow Al + Si + \beta$ -Al₅FeSi eutectic reactions occurred just before or during isothermal holding treatment at 570 °C in the unmodified Al-7Si-0.45Mg sample. Therefore, the unmodified Al-7Si-0.45Mg sample was within the Al-Si eutectic reaction temperature interval during isothermal holding at 570 °C.

The microstructure of the unmodified Al-7Si-0.3Mg alloy isothermally held at 560 °C consisted of α -Al phase, eutectic silicon and a minor fraction of β -Al₅FeSi phase, as shown in Figures 4f and 5e–h. No evidence of the existence of quenched liquid nor of coarse iron-rich intermetallic phases was found after isothermal holding at 560 °C of the unmodified Al-7Si-0.3Mg sample. Therefore, the isothermal holding treatment at 560 °C was carried out with the unmodified Al-7Si-0.3Mg sample in the fully solidified state. Both unmodified Al-7Si-0.45Mg and Al-7Si-0.6Mg samples showed a coarse β -Al₅FeSi phase after isothermal holding treatment at 560 °C. Additionally, the unmodified Al-7Si-0.6Mg sample showed evidence of a minor fraction of quenched liquid after isothermal holding at 560 °C, as shown in Figure 4r. Therefore, the unmodified Al-7Si-0.45Mg and Al-7Si-0.6Mg samples were likely in the liquid-solid state during at least part of the isothermal holding at 560 °C. The microstructure of the quenched liquid regions of the unmodified Al-7Si-0.6Mg alloy isothermally held at 560 °C consisted of fine Al-Si eutectic, π -Al₈FeMg₃Si₆ and Mg₂Si phases, as shown in Figure 5u–x. Therefore, not all magnesium was in α -Al solid solution in the unmodified Al-7Si-0.6Mg alloy after isothermal holding treatment at 560 °C.

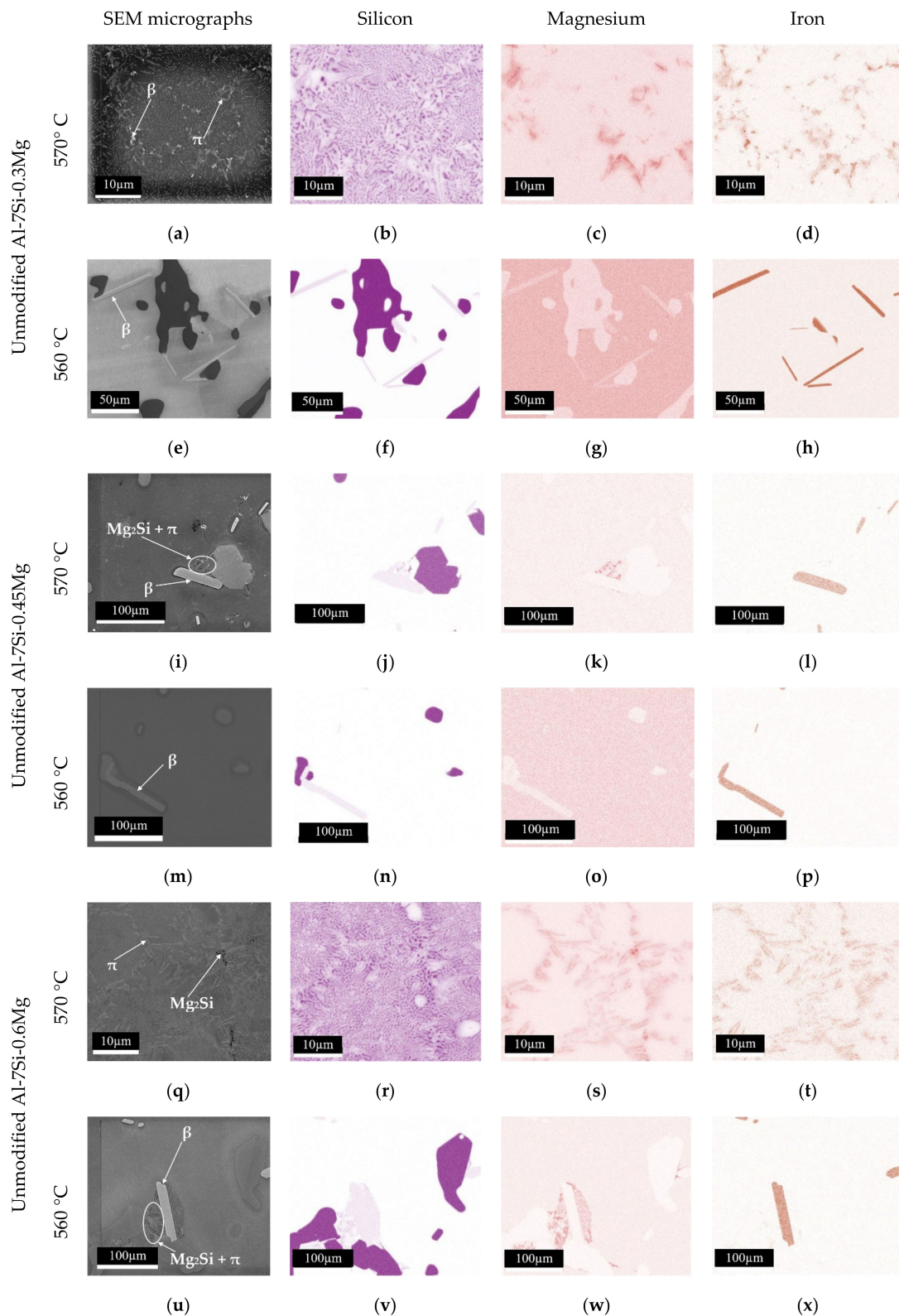


Figure 5. Scanning Electron Microscope (SEM) micrographs and respective Energy-dispersive X-ray Spectroscopy (EDS) mapping showing silicon, magnesium and iron distribution of unmodified Al-7Si-Mg alloys, isothermally held at 560 °C and 570 °C: (a–h) Al-7Si-0.3Mg, (i–p) Al-7Si-0.45Mg, (q–x) Al-7Si-0.6Mg (Color figure online).

3.2. Solute Concentrations in Primary α -Al Solid-Solution after Isothermal Holding Treatments

Figure 6 shows the silicon and magnesium concentrations measured in the interior of the primary α -Al after the different isothermal holding treatments. Additionally, the variations of silicon and magnesium concentrations in the interior of α -Al, and the fractions of eutectic silicon and β -Al₅FeSi formed as a function of temperature predicted by Thermocalc™ (Version 2019a, Thermo-Calc Software, Stockholm, Sweden), are also shown in Figure 6. No significant differences in magnesium and silicon solubility in α -Al as a function of temperature is predicted by Thermocalc™ for both commercial and unmodified Al-7Si-Mg alloys with similar magnesium contents shown in Table 1. For this reason, a base composition of Al-7Si-Mg-0.14Fe was used for the Thermocalc™ calculations shown in Figure 6. It is important to note that both π -Al₈FeMg₃Si₆ and Mg₂Si phases do not form during solidification in equilibrium conditions for the range of compositions used in this study, according to Thermocalc™. The Thermocalc™ calculations show that the silicon concentration in the interior of α -Al increases with a decrease of temperature from 605 °C to the start of the Al-Si eutectic reaction, as seen in Figure 6. The variation of silicon concentration measured in the interior of the primary α -Al phase as a function of the isothermal holding temperature shows a good correlation with the Thermocalc™ calculations for all alloys, as seen in Figure 6. At temperatures above the start of the Al-Si eutectic reaction, the silicon concentration increased with a decrease of the isothermal holding treatment temperature, except for the Al-7Si-0.6Mg sample isothermally held at 595 °C, which showed similar silicon concentration compared to 605 °C. When the samples were isothermally held at 570 °C, i.e., a temperature lower than the start of the Al-Si eutectic reaction according to Thermocalc™, the silicon concentration in the interior of the α -Al globules showed no change, or decreased compared to the samples isothermally held at 575 °C. The results did not show a significant influence of the magnesium content of the alloy on the silicon concentration measured in the interior of α -Al for each isothermal holding temperature. However, the silicon concentrations measured in the interior of the primary α -Al in this study are generally significantly higher than those calculated by Thermocalc™, particularly for temperatures close to/within the Al-Si eutectic reaction. Silicon concentrations up to 1.7 wt.% in the interior of α -Al dendrites have been reported in [11,12]. In this study, silicon concentrations in the interior of α -Al are ≥ 2 wt.% after the isothermal holding treatments at 575 °C for all samples. Figure 4 shows that entrapped eutectic in primary α -Al remains after isothermal holding treatments at 575 °C in all alloys, even for such long time of isothermal holding. The large silicon concentrations measured in the interior of α -Al in this study, particularly for the alloys isothermal held at 575 °C, may result from the influence of entrapped eutectic in the interior of primary α -Al phase on the measurements.

The magnesium concentration in the interior of the primary α -Al phase after isothermal holding treatments increased with an increase of the magnesium content of the alloys, as seen in Figure 6. This result correlates well with the Thermocalc™ calculations, as shown in Figure 6 and the literature [1,23]. At 560 °C, the magnesium concentration in the interior of primary α -Al increases with an increase of the magnesium content of the alloys, as shown in Figure 6 for both Thermocalc™ calculations and measurements. No significant variation of magnesium concentration in the interior of the primary α -Al globules was found in both Al-7Si-0.3Mg and Al-7Si-0.45Mg samples with the decrease of the isothermal holding temperature from 605 to 575 °C. However, for the Al-7Si-0.6Mg sample the magnesium concentration in the interior of α -Al increased with the decrease of the isothermal holding temperature, as predicted by Thermocalc™. The magnesium concentrations in the interior of α -Al of the Al-7Si-Mg samples isothermally held at temperatures above the start of the Al-7Si eutectic reaction were lower compared to the Thermocalc™ calculations for the same temperature range, as seen in Figure 6. Similar results were reported by Sjölander and Seifeddine [13], in which magnesium concentrations measured in the interior of α -Al dendrites of solution treated Al-7Si-Mg castings were lower compared to thermodynamic calculations. However, in this study the samples isothermally held at temperatures above the start of the Al-Si eutectic reaction were in the liquid-solid region while Sjölander and Seifeddine [13] were in the solid-state during solution treatment. Therefore, these results cannot be explained by the magnesium retained in π -Al₈FeMg₃Si₆ as in [13].

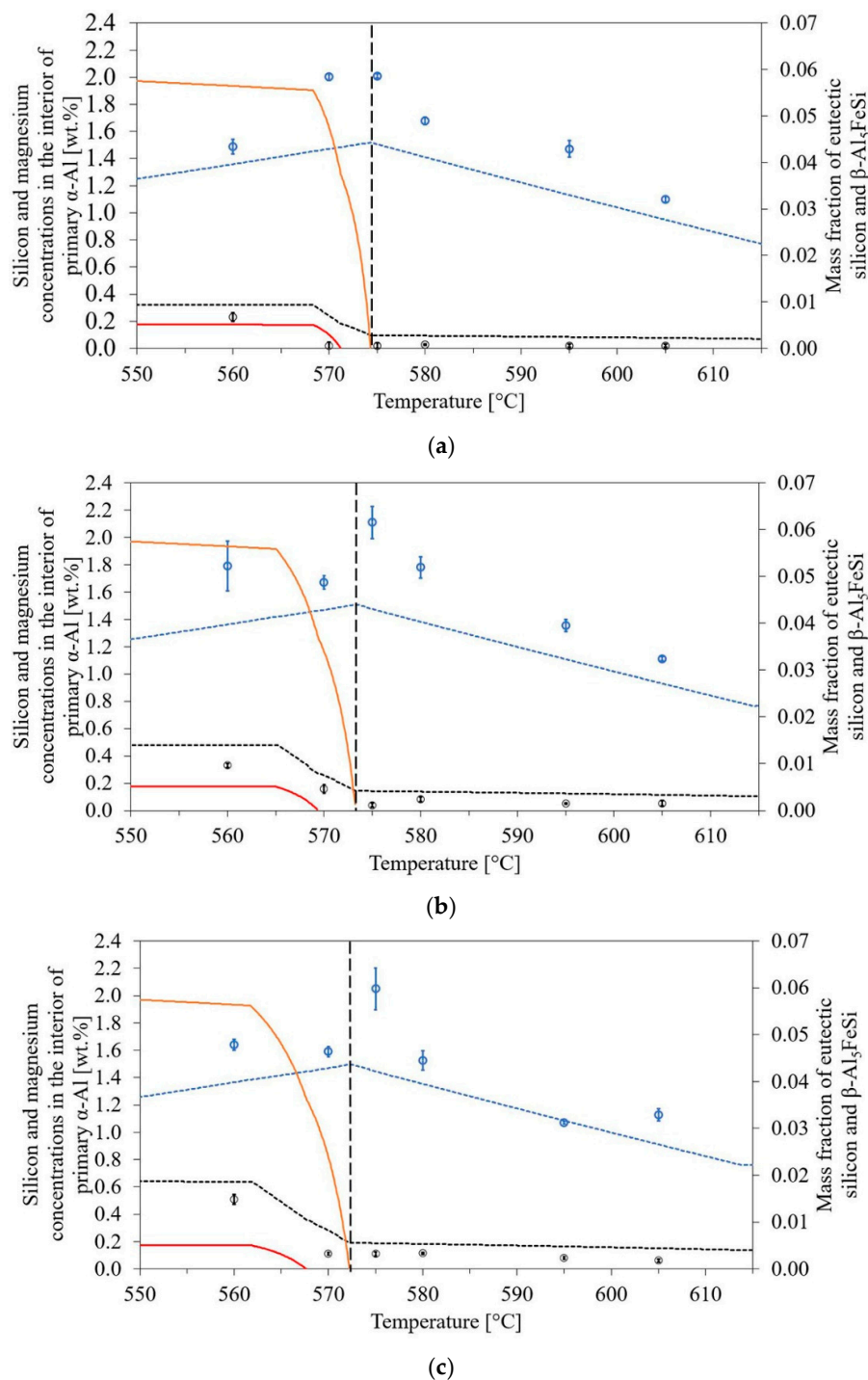


Figure 6. Concentrations of magnesium and silicon measured in the interior of primary α -Al after isothermal holding treatments, and predicted by Thermocalc™ as a function of temperature for the alloys: (a) Al-7Si-0.3Mg, (b) Al-7Si-0.45Mg and (c) Al-7Si-0.6Mg. The vertical dashed lines denote the equilibrium start temperature of the Al-Si eutectic reaction. Error bars show the standard deviation of the measurements.

Figure 6b shows that, after isothermal holding at 570 °C, the magnesium concentration in the interior of primary α -Al of the Al-7Si-0.45Mg sample increased markedly, compared to that obtained after isothermal holding treatments at higher temperatures. Additionally, with the decrease of the isothermal holding temperature from 570 to 560 °C for the Al-7Si-0.45Mg samples, the magnesium concentration increased steeply from 0.16 ± 0.03 to 0.33 ± 0.02 wt.% in the interior of primary α -Al. This result correlates well with the Thermocalc™ calculations. Similarly, a notable increase of the magnesium concentration in the interior of primary α -Al occurs when the isothermal holding treatment temperature is decreased from 570 °C to 560 °C for the Al-7Si-0.3Mg and Al-7Si-0.6Mg samples, as shown in Figure 6a,c. The microstructures obtained after isothermal holding treatments of the Al-7Si-0.6Mg samples suggest that at 570 °C the Al-Si eutectic has not formed while most of the Al-Si eutectic has formed at 560 °C, as seen in Figure 4q,r. After isothermal holding treatments at 560 °C, the magnesium concentration in the interior of primary α -Al was lower than the Thermocalc™ calculations for all alloys. One possible explanation for this result is that the magnesium-rich phases, particularly the π -Al₈FeMg₃Si₆ phase, formed previously during solidification did not dissolve, keeping some magnesium retained, as shown in Figure 5h and reported in other studies [23]. Figure 5f–h,u–x show evidence of π -Al₈FeMg₃Si₆ phase remaining in the Al-7Si-0.3Mg sample isothermally held at 560 °C. The Al-7Si-0.6Mg sample was in the liquid-solid state during isothermal holding treatment at 560 °C, and, most likely, no magnesium-rich phases existed or formed during the isothermal holding treatment. However, both π -Al₈FeMg₃Si₆ and Mg₂Si phases were observed in the quenched liquid regions, as shown in Figure 5u. No evidence of quenched liquid was observed in the Al-7Si-0.45Mg sample isothermally held at 560 °C, which suggests that this sample was in the solid-state at least in the last stages of the isothermal holding treatment that preceded quenching.

3.3. SSM Castings

Representative microstructures of the SSM Al-7Si-0.3Mg, Al-7Si-0.45Mg and Al-7Si-0.6Mg castings are shown in Figure 7. Figure 8 shows SEM micrographs and the respective EDS mapping, with the distribution of silicon, magnesium and iron in the eutectic regions of the SSM Al-7Si-Mg castings, in the as-cast and heat treated to T5 and T6 conditions. The microstructure of the SSM Al-7Si-0.3Mg castings in the as-cast and T5 conditions consisted of primary α -Al globules, modified Al-Si eutectic, and a minor fraction of π -Al₈FeMg₃Si₆ phase in the eutectic regions, as shown in Figures 7a–d and 8a–h. No β -Al₅FeSi plates were detected in SSM Al-7Si-0.3Mg castings in both the as-cast and T5 conditions. Increasing the magnesium content of the alloy did not significantly change the microstructure, except for the formation of script-like Mg₂Si phase, in addition to the π -Al₈FeMg₃Si₆ phase for both SSM Al-7Si-0.45Mg and Al-7Si-0.6Mg castings in the T5 condition, as shown in Figures 7e–h and 8i–p.

Figure 9 shows representative microstructures of the SSM Al-7Si-0.3Mg and Al-7Si-0.45Mg castings in the T6 condition, consisting of primary α -Al globules, spheroidized eutectic silicon and a minor fraction of iron-rich intermetallic phases in the eutectic regions. Fine β -Al₅FeSi plates are observed in the SSM Al-7Si-0.3Mg casting in the T6 condition, and no π -Al₈FeMg₃Si₆ phase was detected. Therefore, most of the π -Al₈FeMg₃Si₆ phase detected in the as-cast SSM Al-7Si-0.3Mg castings transformed into β -Al₅FeSi during solution treatment. A few fine β -Al₅FeSi plates were observed in close association with coarser π -Al₈FeMg₃Si₆ plates in the eutectic regions of the SSM Al-7Si-0.45Mg in the T6 condition. The presence of π -Al₈FeMg₃Si₆ phase in the microstructure of the SSM Al-7Si-0.45Mg casting in the T6 condition suggests that the solution treatment was ineffective, and that some magnesium is retained in the π -Al₈FeMg₃Si₆ phase. The incomplete dissolution of π -Al₈FeMg₃Si₆ phase during solution treatment of the Al-7Si-Mg castings with magnesium contents above 0.4 wt.% was previously reported in [23]. No Mg₂Si phase was detected in both SSM Al-7Si-0.3Mg and Al-7Si-0.45Mg castings in the T6 condition, which suggests that the Mg₂Si phase was completely dissolved, as described in other studies [13,23].

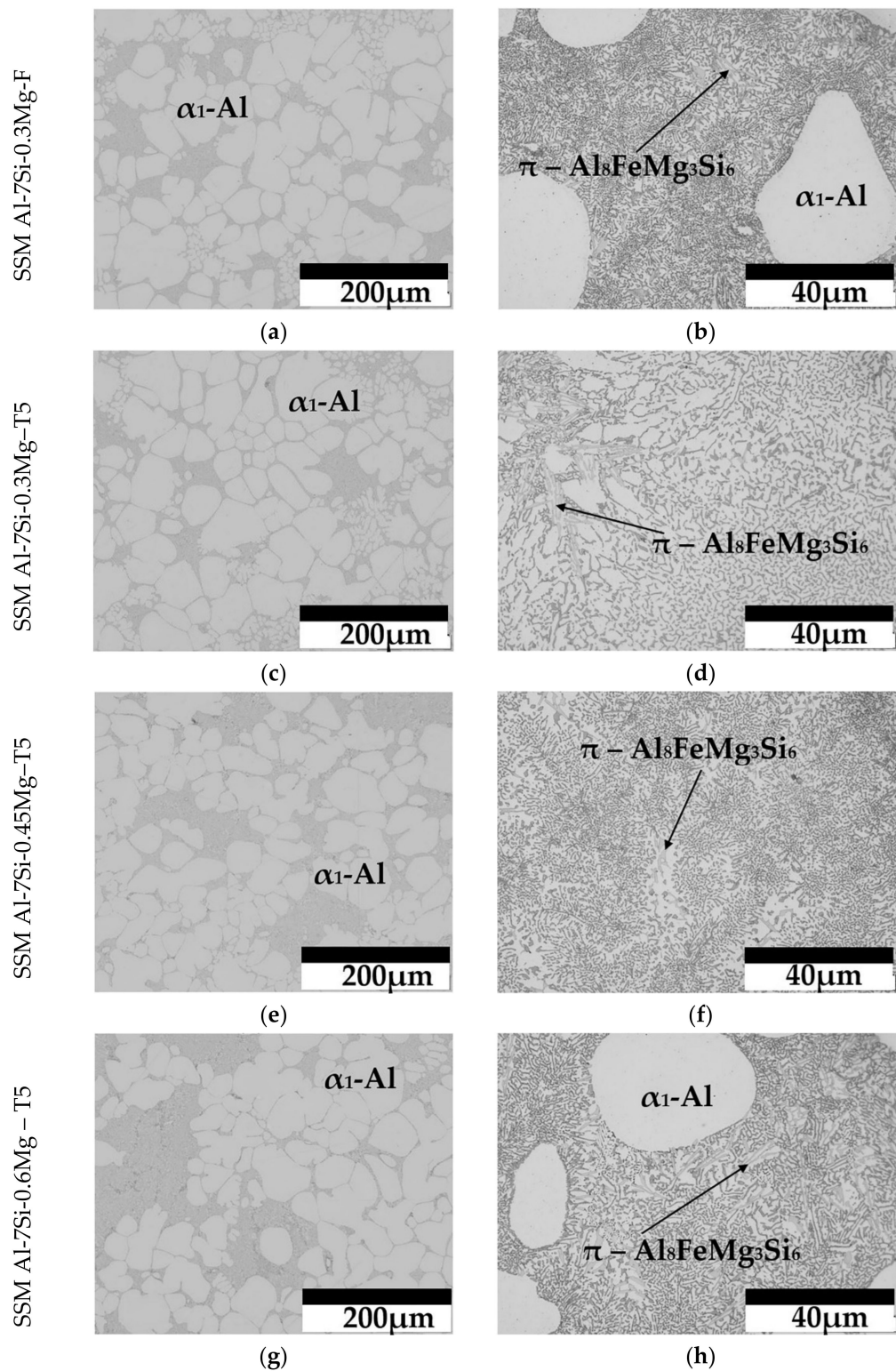


Figure 7. Micrographs of the different Semi-Solid Metal (SSM) castings; (a,b) SSM Al-7Si-0.3Mg-F, (c,d) SSM Al-7Si-0.3Mg-T5, (e,f) SSM Al-7Si-0.45Mg-T5, (g,h) SSM Al-7Si-0.6Mg-T5. Light grey: iron-rich intermetallics. Dark grey: eutectic silicon.

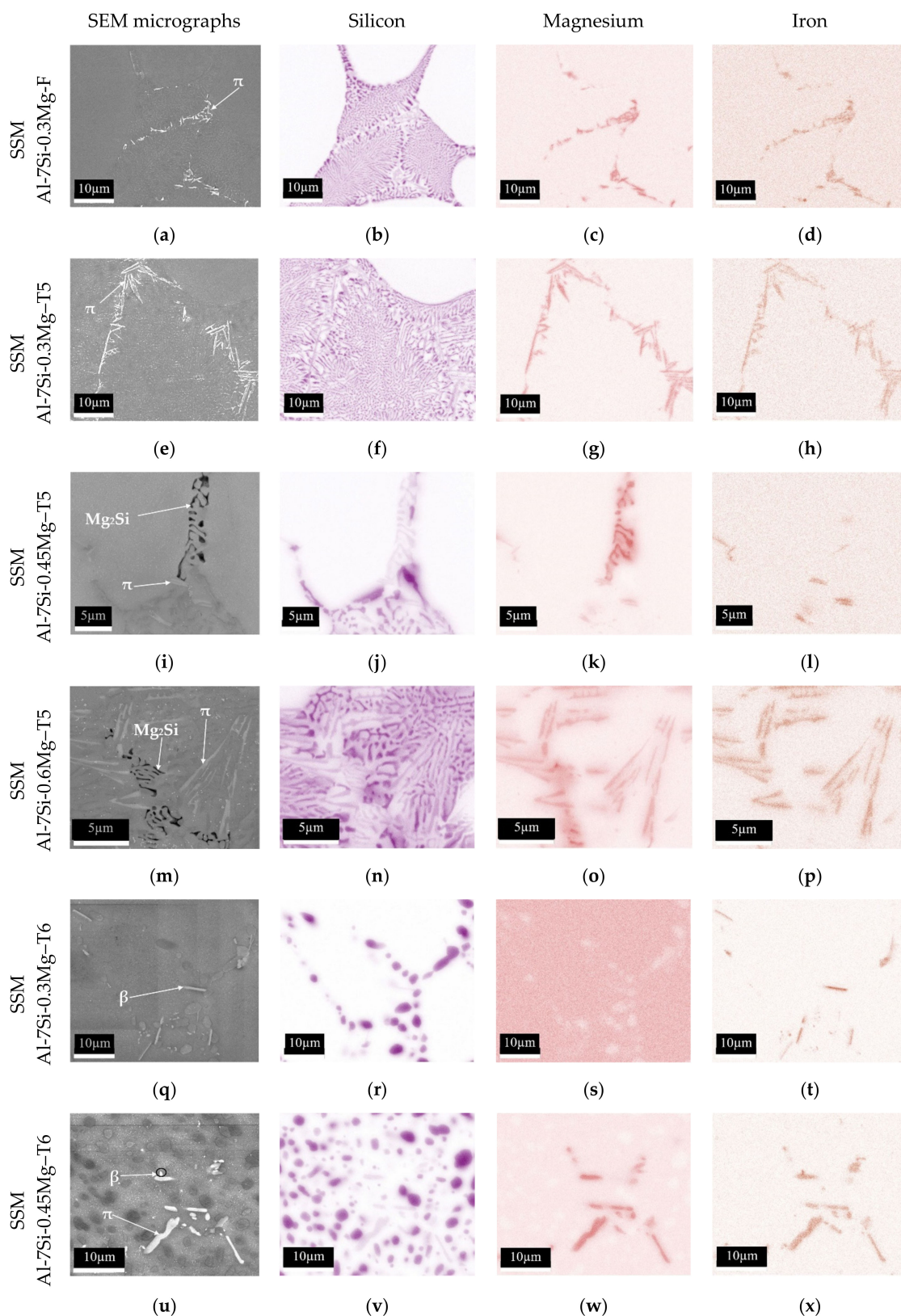


Figure 8. SEM micrographs and respective EDS mapping showing silicon, magnesium and iron distribution in SSM Al-7Si-Mg castings: (a–d) SSM Al-7Si-0.3Mg-F, (e–h) SSM Al-7Si-0.3Mg-T5, (i–l) SSM Al-7Si-0.45Mg-T5, (m–p) SSM Al-7Si-0.6Mg-T5, (q–t) SSM Al-7Si-0.3Mg-T6 and (u–x) SSM Al-7Si-0.45Mg-T6 (Color figure online).

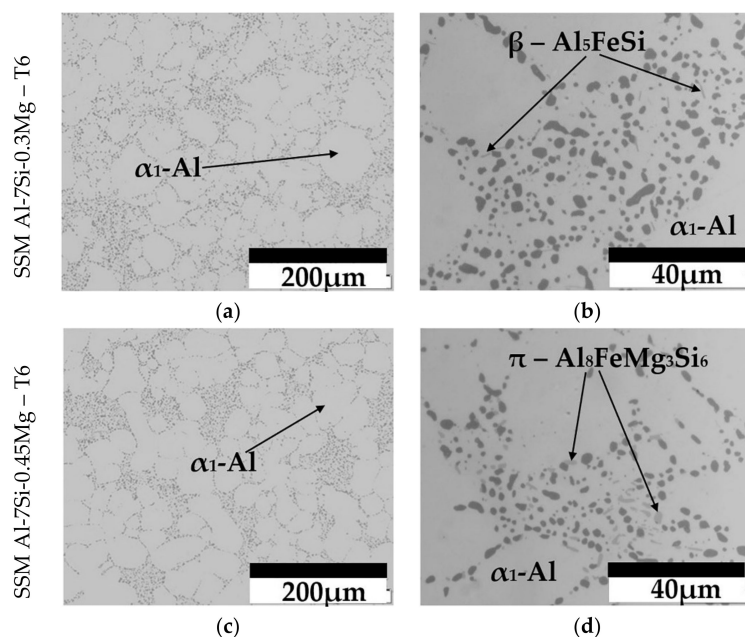


Figure 9. Micrographs of the different SSM castings: (a,b) SSM Al-7Si-0.3Mg—T6, (c,d) SSM Al-7Si-0.45Mg—T6. Light grey: iron-rich intermetallics. Dark grey: eutectic silicon.

3.4. Silicon and Magnesium Concentrations in the Interior of Primary α_1 -Al Globules of SSM Castings

Two populations of primary α -Al globules were distinguished in the microstructures of SSM Al-7Si-Mg castings. The so-called α_1 -Al globules that formed during slurry preparation, or in the shot sleeve with a diameter $\geq 25 \mu\text{m}$ and in-cavity solidified grains that have a diameter $< 25 \mu\text{m}$ [25]. Figure 10 shows silicon and magnesium concentrations in the interior of α_1 -Al globules of SSM Al-7Si-Mg castings in the different conditions. The silicon and magnesium concentrations were measured in the interior of α_1 -Al globules, because these globules most likely have a greater influence on mechanical properties, compared to the smaller in-cavity solidified crystals. The results showed no significant differences in the silicon concentration in the interior of α_1 -Al globules of the different SSM castings. Only the SSM Al-7Si-0.45Mg casting in the T5 condition has a slightly higher silicon concentration, compared to the other castings, but still not significant.

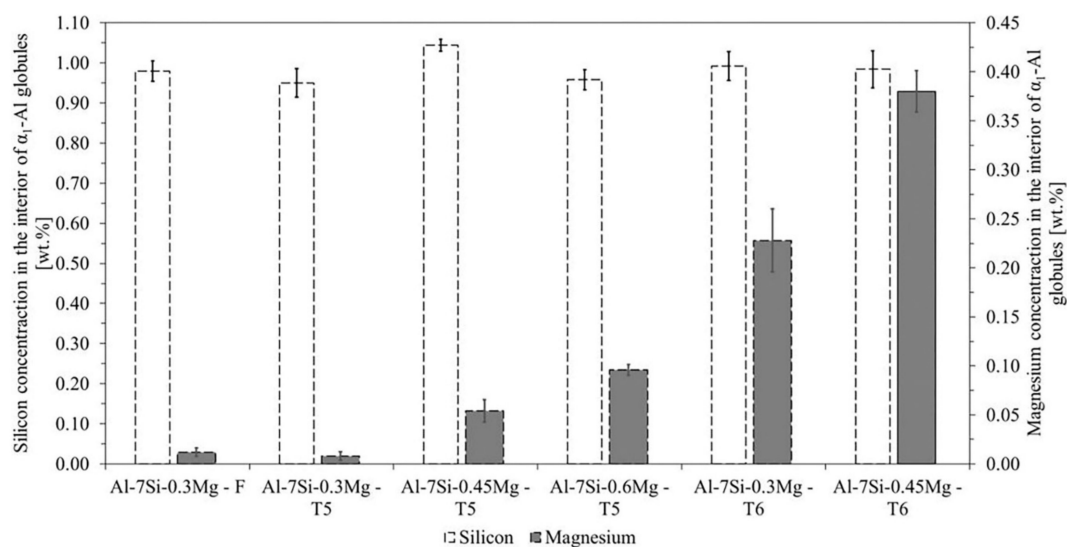


Figure 10. Silicon and magnesium concentrations in the interior of α_1 -Al globules obtained for the different SSM castings. Error bars show the standard deviation of the measurements.

The magnesium concentration in the interior of α_1 -Al globules was very low for all SSM Al-7Si-Mg castings in the as-cast and T5 conditions, compared to the magnesium content of the commercial alloys. However, after T6 heat treatment, the magnesium concentration in the interior of α_1 -Al globules increased markedly, compared to the SSM castings in as-cast and T5 conditions, as shown in Figure 10. The increase of the magnesium content of the commercial alloy resulted in an increase of the magnesium concentration in the interior of the α_1 -Al globules of the SSM castings in both T5 and T6 conditions, similar to the isothermal holding treatment results, as shown in Figure 6.

3.5. Mechanical Properties

Figure 11 shows the 0.2% offset yield strength, tensile strength and % elongation of the SSM Al-7Si-Mg castings in the different conditions. The 0.2% offset yield strength increased with an increase of magnesium content of the castings for both T5 and T6 conditions, as reported in other studies [15,16]. Additionally, the 0.2% offset yield strength of the SSM Al-7Si-Mg castings in the T6 condition is significantly higher compared to the SSM Al-7Si-Mg castings in both as-cast and T5 conditions, as shown in Figure 11. The 0.2% offset yield strength of the SSM Al-7Si-0.3Mg casting increased slightly after T5 heat treatment compared to the as-cast condition. Therefore, precipitation hardening occurred in the SSM Al-7Si-0.3Mg castings in the T5 condition, even for such a low magnesium concentration measured in the interior of α_1 -Al globules for these castings, as shown in Figure 10. However, the precipitation hardening during aging to T5 condition is significantly lower compared to that obtained after T6 heat treatment.

The tensile strength of the SSM Al-7Si-0.3Mg castings in both as-cast and T5 conditions was similar. No significant differences in tensile strength was found between SSM Al-7Si-0.3Mg and Al-7Si-0.45Mg castings in the T5 condition. However, the SSM Al-7Si-0.6Mg casting in the T5 condition showed higher tensile strength, compared to the other SSM castings in the same condition. After T6 heat treatment, the tensile strength of the castings increased in comparison to both as-cast and T5 conditions.

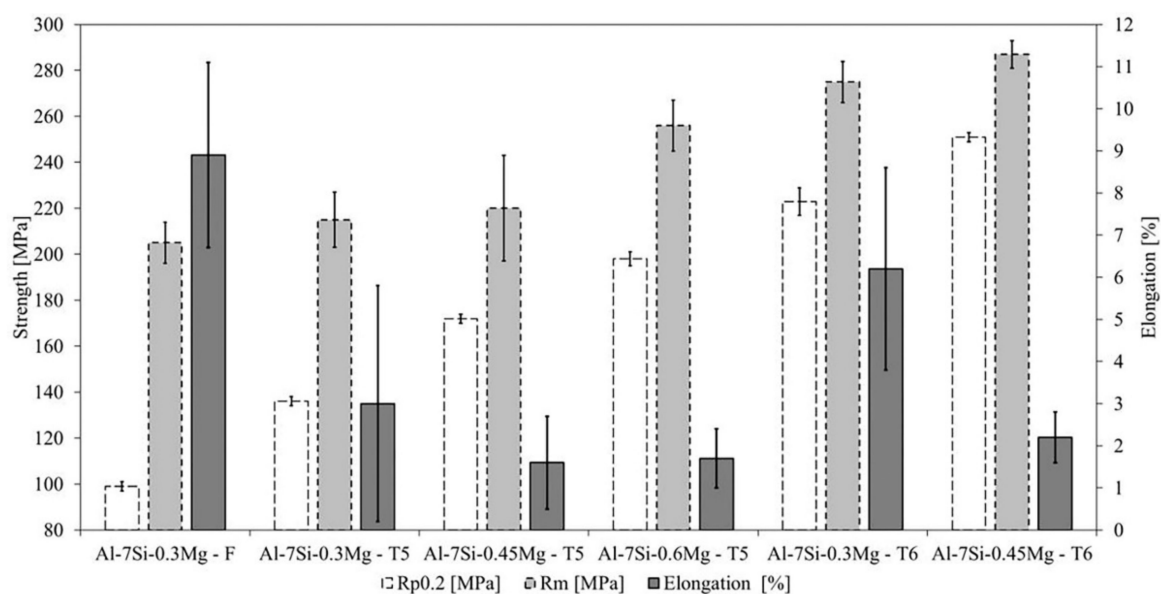


Figure 11. Tensile strength, 0.2% offset yield strength and % elongation of SSM castings in the different conditions. Error bars show the standard deviation of the measurements.

After T5 heat treatment, the elongation obtained for the SSM Al-7Si-0.3Mg castings decreased in comparison to the as-cast condition due to precipitation hardening [26]. The increase of magnesium content of the commercial alloy did not affect the elongation of the castings in the T5 condition, while, for the castings in the T6 condition, the increase of magnesium content resulted in a decrease of elongation, as reported in other studies [26]. Additionally, the elongation of the SSM Al-7Si-0.3Mg

casting in the T6 condition was higher compared to the SSM Al-7Si-0.3Mg casting in the T5 condition, as reported in the literature [10]. However, for the SSM Al-7Si-0.45Mg casting, a similar elongation was obtained in both the T5 and T6 conditions. Figure 8u shows that most of π -Al₈FeMg₃Si₆ phase still remained in the microstructure of the SSM Al-7Si-0.45Mg casting after T6 heat treatment, which may explain the low elongation obtained. A larger scatter in the elongation values was observed for the SSM Al-7Si-0.3Mg castings, compared to the castings with higher magnesium contents, as shown in Figure 11. This is an indication of more defects obtained in the SSM Al-7Si-0.3Mg castings compared to the other castings.

4. Discussion

4.1. Variation of Solute Concentration in the Interior of the Primary α -Al Phase after Isothermal Holding Treatments

The results of this study suggest that the formation of eutectic silicon and β -Al₅FeSi phases during eutectic solidification of Al-7Si-Mg samples generate a sharp increase in the magnesium concentration in the interior of primary α -Al. No evidence of coarse eutectic silicon or intermetallic phases were found in the microstructure of the different Al-7Si-Mg samples isothermally held at temperatures higher than 570 °C, as shown in Figures 3 and 4. Therefore, the Al-Si eutectic reaction most likely did not occur for isothermal holding at temperatures above 570 °C. After isothermal holding treatment at 570 °C, the Al-7Si-0.45Mg sample showed two distinct eutectic microstructures, as seen in Figure 4h,k; a coarse eutectic microstructure that consisted of coarse Al-Si eutectic and β -Al₅FeSi that existed/formed during the isothermal holding at 570 °C, and a much finer eutectic microstructure solidified during quenching. Therefore, most likely, both $L \rightarrow Al + Si$ and $L \rightarrow Al + Si + \beta$ -Al₅FeSi eutectic reactions occurred to some extent during isothermal holding at 570 °C for the Al-7Si-0.45Mg sample. Simultaneously, a steep increase of the magnesium concentration in the interior of primary α -Al in the Al-7Si-0.45Mg sample occurred during isothermal holding treatment at 570 °C compared to higher temperatures. During isothermal holding at 570 °C of the Al-7Si-0.45Mg sample, the silicon in the existing liquid of the Al-7Si-0.45Mg sample was partly consumed to form eutectic silicon (100% Si) and β -Al₅FeSi phases (15% Si) [27]. Consequently, the ratio wt.%Mg/wt.%Si in the existing liquid increases with the formation of eutectic silicon and β -Al₅FeSi phase during isothermal holding at 570 °C for the Al-7Si-0.45Mg sample. Additionally, the concentration of silicon in the interior of primary α -Al decreases for the Al-7Si-0.45Mg sample after isothermal holding at 570 °C, compared to isothermal holding at 575 °C, as shown in Figure 6b. Dons et al. [12] showed that in slowly cooled Al-7Si alloys, the lower concentration of silicon at the edges of primary α -Al dendrite arms compared to its centre results from the growth of eutectic silicon after solidification [12]. Therefore, it is reasonable to assume that the decrease of silicon and increase of magnesium concentrations in the interior of primary α -Al of the Al-7Si-0.45Mg sample are generated by the formation and growth of eutectic silicon during isothermal holding treatment at 570 °C.

For both unmodified Al-7Si-0.3Mg and Al-7Si-0.6Mg alloys, no coarse eutectic silicon or β -Al₅FeSi phases were detected after isothermal holding treatment at 570 °C, which suggests that the Al-Si eutectic reaction has not occurred at this temperature. Consequently, magnesium concentration in the interior of primary α -Al did not change significantly for both Al-7Si-0.3Mg and Al-7Si-0.6Mg samples isothermal held at 570 °C, compared to 575 °C, as shown in Figure 6a,c. During the isothermal holding treatment at 560 °C, the Al-7Si-0.3Mg sample was in the solid-state while the Al-7Si-0.6Mg sample contained a minor fraction of liquid. These observations are supported by microstructural analysis that showed quenched liquid in some regions of the Al-7Si-0.6Mg sample isothermal held at 560 °C, but not in the Al-7Si-0.3Mg sample, as shown in Figure 4l,r. No regions of quenched liquid were found in the microstructure of the Al-7Si-0.45Mg sample isothermally held at 560 °C, which suggests that this sample was in the solid-state at least in the moments that preceded quenching. However, it is possible that the Al-7Si-0.45Mg sample was in the liquid-solid state during some period of the isothermal holding treatment at 560 °C. The Al-7Si-0.3Mg sample isothermally held at 560 °C showed

evidence of $\pi\text{-Al}_8\text{FeMg}_3\text{Si}_6$ phase remained in the microstructure. Therefore, it would be expected that $\pi\text{-Al}_8\text{FeMg}_3\text{Si}_6$ phase could be found at greater extent in the Al-7Si-0.45Mg sample isothermally held at 560 °C, due to the higher magnesium content compared to the Al-7Si-0.3Mg sample [23]. One possible explanation for this result is that the Al-7Si-0.45Mg sample was in the liquid-solid state from the start-up to a specific extension of the isothermal holding treatment at 560 °C. This would imply that eutectic silicon and $\beta\text{-Al}_5\text{FeSi}$ phases could form and grow larger in the existing liquid during the isothermal holding. Large $\beta\text{-Al}_5\text{FeSi}$ phase is observed in the microstructure of the Al-7Si-0.45Mg sample isothermally held at 560 °C, similar to that obtained in the Al-7Si-0.6Mg sample isothermally held at the same temperature, as shown in Figure 5m,u. However, magnesium concentration in the interior of primary $\alpha\text{-Al}$ was lower compared to that predicted by Thermocalc™, which suggests that $\pi\text{-Al}_8\text{FeMg}_3\text{Si}_6$ phase still remains in the microstructure. The microstructure of the Al-7Si-0.6Mg sample isothermally held at 560 °C shows $\pi\text{-Al}_8\text{FeMg}_3\text{Si}_6$ phase only in the quenched liquid regions, which suggests that the $\pi\text{-Al}_8\text{FeMg}_3\text{Si}_6$ and Mg_2Si phases only formed in the very last solidified liquid. This indicates that the magnesium concentration that remained in the existing liquid of the Al-7Si-0.6Mg sample during isothermal holding at 560 °C was enough to promote the quaternary eutectic reaction $L \rightarrow \text{Al} + \text{Si} + \pi\text{-Al}_8\text{FeMg}_3\text{Si}_6 + \text{Mg}_2\text{Si}$ during quenching, as seen in Figure 5u-x. Therefore, the formation of the $\pi\text{-Al}_8\text{FeMg}_3\text{Si}_6$ phase can be prevented only when all the remaining liquid solidifies under equilibrium conditions for Al-7Si-0.6Mg alloys.

At the start of the isothermal holding treatment at 560 °C, the magnesium concentration in the interior of primary $\alpha\text{-Al}$ of the Al-7Si-0.45Mg sample is most likely significantly lower, compared to that obtained just before quenching. This is supported by the fact that it is highly improbable that the steep increase of the magnesium concentration, in the interior of primary $\alpha\text{-Al}$ in the Al-7Si-0.45Mg sample, isothermally held at 560 °C was generated by the cooling rate in the furnace, compared to that obtained at 570 °C. During isothermal holding treatment of the Al-7Si-0.45Mg sample at 560 °C, the formation and growth of the eutectic silicon and $\beta\text{-Al}_5\text{FeSi}$ phases generate an increase of the magnesium concentration in the interior of primary $\alpha\text{-Al}$ and consequently, a depletion of magnesium of the remaining liquid. Therefore, the solidus temperature of the remained liquid is increased as the magnesium concentration decreases [23]. The formation of the eutectic Al-Si and $\beta\text{-Al}_5\text{FeSi}$ phases proceed in the remaining liquid until the solidification is complete during isothermal holding at 560 °C of the Al-7Si-0.45Mg sample. It is likely that some of $\pi\text{-Al}_8\text{FeMg}_3\text{Si}_6$ or Mg_2Si phases remained in the Al-7Si-0.45Mg sample after isothermal holding treatment at 560 °C, because of the lower magnesium concentration in the interior of primary $\alpha\text{-Al}$, compared to the concentration predicted by Thermocalc™, as shown in Figure 6b.

An increase of approximately 12-, 10- and 5-fold of magnesium concentration in the interior of primary $\alpha\text{-Al}$ occurred in the Al-7Si-0.3Mg, Al-7Si-0.45Mg and Al-7Si-0.6Mg samples, respectively, after isothermal holding treatment at 560 °C compared to isothermal holding at 575 °C. The increase of the magnesium concentration was similar for Al-7Si-0.45Mg and Al-7Si-0.6Mg samples, most likely due to the higher magnesium content retained in the $\pi\text{-Al}_8\text{FeMg}_3\text{Si}_6$ phase for the Al-7Si-0.6Mg sample. It is interesting to note that the increase of the magnesium content of the unmodified alloy generates a less significant increase of magnesium concentration in the interior of primary $\alpha\text{-Al}$ during the Al-Si eutectic reaction, as shown in Figure 6. One possible explanation for this result is that the increase of magnesium content of the alloy generates a higher increase of the magnesium concentration in the interior of primary $\alpha\text{-Al}$ at temperatures higher than the Al-Si eutectic reaction.

4.2. Silicon and Magnesium Concentrations in the Interior of Primary $\alpha_1\text{-Al}$ of SSM Castings

During the Rheometal™ slurry preparation process, primary $\alpha_1\text{-Al}$ crystals nucleate and grow dispersed in the solute enriched liquid. Subsequently, the prepared slurry is poured into the shot sleeve and injected into the die cavity, where the solidification proceeds with the growth of $\alpha_1\text{-Al}$ crystals from the slurry and formed in the shot sleeve, nucleation and growth of in-cavity solidified crystals and eutectic. The solute concentration in the interior of primary $\alpha_1\text{-Al}$ globules at the end of the slurry

preparation process is likely similar to that obtained in the interior of α_1 -Al globules in the as-solidified SSM castings, as shown in [25]. This can be explained by the short time that the slurry is held in the shot sleeve and the high cooling rate during solidification in the die-cavity, which prevents extensive solid-state diffusion. For the SSM Al-7Si-Mg castings in the as-cast and T5 conditions, the silicon concentration in the interior of primary α_1 -Al globules is mostly determined by the temperature of the slurry before injection into the die-cavity. The commercial alloys used to produce the SSM castings contained similar silicon contents, and the slurry preparation parameters were kept constant in all experiments. Consequently, no significant differences were found in the silicon concentration in the interior of primary α_1 -Al globules for the SSM Al-7Si-Mg castings in both as-cast and T5 conditions. The silicon concentration in the interior of primary α -Al of the SSM castings in the T6 condition was mainly dictated by the temperature at which the solution treatment was carried out, which was kept constant for all SSM castings. No significant differences were found in the silicon concentrations in the interior of primary α -Al for SSM castings in both T6 and T5 conditions. This suggests that the solution treatment temperature used in the T6 heat treatment generated similar silicon concentration in the interior of primary α_1 -Al globules compared to that obtained after slurry preparation, as shown in Figure 6.

Figure 10 shows that the magnesium concentration in the interior of the primary α_1 -Al globules of SSM castings in both as-cast and T5 conditions is significantly lower, compared to that obtained for the castings in the T6 condition. The isothermal holding treatment results showed that at temperatures above the start of the Al-Si eutectic reaction the magnesium concentration in the interior of primary α -Al is significantly lower, compared to lower temperatures, as shown in Figure 6. The average silicon concentration in the interior of primary α_1 -Al globules of each SSM casting in both as-cast and T5 conditions was used to determine the corresponding solidus temperature in Thermocalc™ [25]. The corresponding solidus temperatures obtained were within the range of 605 to 599 °C. Interestingly, the magnesium concentrations in the interior of the α_1 -Al globules of SSM castings in both as-cast and T5 conditions correlate well with that obtained after isothermal holding treatments at 605 °C and 595 °C of the equivalent commercial alloys. During the Rheometal™ slurry preparation process primary α_1 -Al globules are formed and held within the semi-solid range of temperatures up to about 30 s, before injection into the die-cavity. This can promote the homogenization of the α_1 -Al globules composition, as described in [25].

Figure 6 shows that, for equilibrium conditions, the magnesium concentration in the interior of primary α -Al increases substantially during the Al-Si eutectic reaction. However, the high cooling rates in the die-cavity likely limit the time for diffusion in the solid-state to a great extent, and for this reason, the concentrations of magnesium and silicon in the interior of α_1 -Al globules likely remained constant during solidification in the die-cavity. Therefore, the Al-Si eutectic reaction occurs at great undercooling in a liquid highly enriched in silicon and magnesium. The high cooling rate and the composition of the remaining liquid favor the formation of the iron-rich π -Al₈FeMg₃Si₆, instead of the β -Al₅FeSi phase, during the eutectic reaction of SSM Al-7Si-Mg castings, as shown in Figure 8. For the as-solidified SSM Al-7Si-0.45Mg and SSM Al-7Si-0.6Mg castings, the Mg₂Si phase forms in addition to π -Al₈FeMg₃Si₆, most likely due to the higher magnesium content in the remaining liquid, compared to that of the SSM Al-7Si-0.3Mg casting.

During solution treatment, the casting is held at a temperature lower than Al-Si eutectic reaction to homogenize the primary α -Al, dissolve π -Al₈FeMg₃Si₆ and Mg₂Si phases, releasing silicon and magnesium into α -Al solid-solution, and spheroidize eutectic silicon [13]. The results of isothermal holding treatments showed that the magnesium concentration in the interior of primary α -Al increases significantly for the Al-7Si-Mg samples held at temperatures below the start of the Al-Si eutectic reaction. Therefore, the magnesium concentration in the interior of the primary α_1 -Al globules of SSM castings in the T6 condition is markedly higher, compared to the castings in both as-cast and T5 conditions. In the SSM Al-7Si-0.3Mg casting in the T6 condition, β -Al₅FeSi phase was the only iron-rich phase detected, which suggests that most of the π -Al₈FeMg₃Si₆ phase formed during solidification

transformed into β -Al₅FeSi and magnesium was released into the α -Al solid-solution during solution treatment. However, for the SSM Al-7Si-0.45Mg casting in the T6 condition the transformation of π -Al₈FeMg₃Si₆ phase into β -Al₅FeSi was incomplete, because both phases were detected in the microstructure, as shown in Figure 8u–x. The magnesium retained in π -Al₈FeMg₃Si₆ phase will not contribute to precipitation hardening during aging. The incomplete dissolution of the π -Al₈FeMg₃Si₆ phase during solution treatment occurs typically in Al-7Si-Mg castings with magnesium contents higher than 0.4 wt.% [17,23].

4.3. Mechanical Properties

The low magnesium concentrations in the interior of α_1 -Al globules of as-solidified SSM Al-7Si-0.3Mg castings result in a limited fraction of precipitates formed during subsequently aging. Therefore, just a slight increase of strength occurs for the SSM Al-7Si-0.3Mg casting in the T5 condition, compared to the SSM Al-7Si-0.3Mg casting in the as-cast condition, as shown in Figure 11. For the SSM castings in the T5 and T6 conditions the 0.2% offset yield strength increased with the increase of magnesium content in the commercial alloy, as reported in other studies [15,16]. The increase of magnesium content in the commercial alloys results in an increase of the magnesium concentration in the interior of the primary α_1 -Al globules and, consequently, the potential for precipitation hardening during aging is improved, as seen in Figures 10 and 11.

No significant differences were found in the silicon concentration in the interior of α_1 -Al globules for the SSM castings in the T5 and T6 conditions. Therefore, the precipitation hardening during artificial aging is mostly governed by the magnesium concentration in the interior of α_1 -Al solid-solution at the start of the artificial aging. Caceres et al. [17] suggested that the 0.2% offset yield strength is proportional to the square root of the magnesium content of the castings in the T6 condition, assuming that all magnesium of the commercial alloy is in α -Al solid-solution. The 0.2% offset yield strength of the Al-7Si-Mg castings in the T5 and T6 conditions was plotted in Figure 12 as a function of the square root of the magnesium concentration in the interior of α_1 -Al globules. It was found that the 0.2% offset yield strength of the Al-7Si-Mg castings in the T5 and T6 conditions in this study is proportional to the square root of the magnesium concentration in the interior of α_1 -Al globules formed during the slurry preparation process.

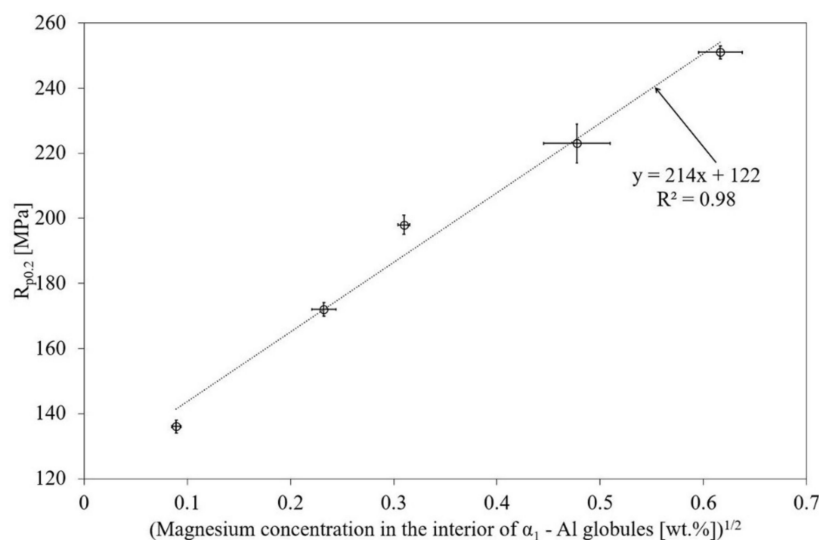


Figure 12. $R_{p0.2}$ as a function of the square root of the magnesium concentration in the interior of α_1 -Al globules of SSM castings in the T5 and T6 conditions. The dashed line shows the best fit for the presented data. Error bars show the standard deviation of the measurements.

Figure 11 shows that the SSM Al-7Si-0.6Mg casting has the highest tensile strength compared to the other castings in the T5 condition, while elongation was similar. For the castings in the T6 condition, the tensile strength increased and elongation decreased with an increase of the magnesium content of the commercial alloy. The lower ductility obtained for the Al-7Si-0.45Mg compared to the Al-7Si-0.3Mg castings in the T6 condition most likely results from the brittle and coarse π -Al₈Mg₃FeSi₆ phase that remained in the Al-7Si-0.45Mg casting [17,23]. The increase in ductility of the SSM Al-7Si-0.3Mg castings in the T6 condition, in comparison to the same commercial alloy castings in the T5 condition, is probably generated by the transformation of the π -Al₈Mg₃FeSi₆ phase into very fine β -Al₅FeSi platelets. However, for the Al-7Si-0.45Mg casting the elongation was similar in both the T6 and T5 conditions. This result can be explained by the π -Al₈Mg₃FeSi₆ platelets that still remained in the microstructure after solution treatment for the Al-7Si-0.45Mg casting, as observed in other studies [17]. It was observed that the Al-7Si-0.3Mg castings show larger scatter in the elongation values in comparison to the Al-7Si-0.45Mg and Al-7Si-0.6Mg castings, as shown in Figure 11. This suggests that more defects formed in the SSM Al-7Si-0.3Mg castings. Additionally, it is expected that both SSM Al-7Si-0.45Mg and Al-7Si-0.6Mg castings have improved precipitation hardening, and formation of π -Al₈Mg₃FeSi₆ and Mg₂Si phases, which can decrease the ductility at such an extent that overlap the effect of defects on ductility.

5. Conclusions

The magnesium and silicon solubilities in primary α -Al phase were studied in Al-7Si-Mg alloys isothermally held at different temperatures in the liquid-solid range for 120 h. The formation and growth of eutectic silicon and β -Al₅FeSi phases during the binary and ternary eutectic reactions generates a decrease of the silicon concentrations in the interior of primary α -Al phase and remaining liquid for long periods of isothermal holding. When solid-state diffusion is allowed, the decrease of silicon concentrations in both the interior of the primary α -Al phase and the remaining liquid seems to trigger a remarkable increase of magnesium concentration in the interior of primary α -Al phase. However, when solid-state diffusion during Al-Si eutectic solidification is reduced to negligible values, due to high cooling rates, no substantial increase of magnesium concentration occurs in the interior of the primary α -Al phase, and magnesium-rich phases are formed in the quenched liquid.

SSM Al-7Si-Mg castings were produced with varying magnesium contents, and heat-treated to T5 and T6 conditions. The high cooling rates during solidification in the die-cavity reduce the solid-state diffusion, to such an extent that most of the magnesium of the alloy is retained in π -Al₈Mg₃FeSi₆ and Mg₂Si phases formed in the solute-enriched remaining liquid. Therefore, the magnesium concentration in the interior of primary α_1 -Al globules is relatively low in the as-solidified Al-7Si-Mg castings, which results in little increase of 0.2% offset yield strength during aging to T5 condition, compared to the as-cast condition. During solution treatment, the π -Al₈Mg₃FeSi₆ phase transforms into β -Al₅FeSi phase at certain extent, which depends on the magnesium content of the alloy, and Mg₂Si is dissolved releasing magnesium into α -Al solid-solution. The magnesium concentration in the interior of α -Al increases substantially during solution treatment and, subsequently, precipitation hardening during aging to T6 condition is markedly improved compared to the T5 condition. Therefore, the mechanical properties obtained for SSM castings in the T5 condition are much lower compared to castings in the T6 condition, mostly due to the high cooling rates during eutectic solidification that reduce solid-state diffusion to such an extent that magnesium remains retained in eutectic phases. The 0.2% offset yield strength of the SSM Al-7Si-Mg castings in the T5 and T6 conditions increase linearly with the square root of the magnesium concentration in the interior of the α_1 -Al globules formed during slurry preparation.

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