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Optimization of heat treatment of gravity cast Sr-modified B356 aluminum alloy

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Abstract: In recent years, certain foundry processes have made it possible to obtain products with very thin parts, below the 4 mm threshold of the permanent mold casting technology. The safety margins of these castings have been reduced, so the T6 heat treatment conditions adopted for the Al−7Si−Mg alloys need to be investigated to identify the best combination of strength and ductility. Furthermore, the cost and the production time associated with T6 heat treatment have to be optimized. In the present work, an experimental study was carried out to optimize the solution treatment and artificial aging conditions in gravity cast thin bars of B356 aluminum alloy modified with Sr. Two solution temperatures were selected, 530 °C and 550 °C, respectively, with solution time ranging from 2 to 8 h, followed by water quenching and artificial aging at 165 °C with aging time from 2 to 32 h. The results of hardness and tensile tests were correlated with differential scanning calorimetry (DSC) analysis. The best combination of mechanical properties and heat treatment duration was obtained with 2 h solutionizing at 550 °C and 8 h aging at 165 °C. DSC analysis showed that the alloy's mechanical properties reach the maximum value when the β'' phase is completely developed during the artificial aging.

Key words: B356 aluminum alloy; gravity casting; heat treatment; hardness; tensile property; β'' phase

1 Introduction

Cast Al−7Si−Mg alloys have been widely used in automotive applications thanks to their high specific strength. Their typical microstructure consists of primary α(Al) grains and eutectic structures. In the as-cast condition, the α(Al) grains are much softer than the eutectic structures and the mechanical properties of the alloy are low. T6 heat treatment is usually used to obtain the desired mechanical properties: a solution treatment, a quenching and an artificial aging strengthen the primary α(Al) grains. The solution treatment dissolves the equilibrium β phase (Mg_2Si particles) in the Al matrix, homogenizes the alloying elements in the casting and modifies the morphology of the eutectic structures. The amounts in solution and the rate of dissolution increase as the solution temperature rises, but this parameter is limited by the melting point of the eutectic phases: complex eutectics (predominately iron-rich particles) melt at temperatures below the equilibrium eutectic temperature. This phenomenon occurs at the grain boundaries and reduces the mechanical properties of the alloy [1]. Solution treatment time is also an important parameter: the time needed for dissolution and homogenization depends on the composition, morphology, size and distribution of the phases present after solidification, as well as the temperature of the solution treatment. Previous works [2,3] showed that the dissolution of Mg and Si is complete in a short solution treatment time and creates large amounts of vacancies and distortions in the Al matrix. These vacancies and distortions need long solution time to be recovered on account of the atomic diffusion. A high quench rate following the solution treatment suppresses the precipitation of the β phase and keeps vacancies and distortions unchanged, forming a supersaturated solid solution. The hardening elements can precipitate from the supersaturated solid solution during the subsequent artificial aging. These precipitates strengthen the alloy, because they act as obstacles to the dislocation motion. Vacancies and distortions facilitate the nucleation of fine precipitates, which strengthen the alloy more than larger ones. The eutectic Si particles also strengthen the alloy, by undergoing the following transformations during solution treatment: fragmentation, spheroidization and

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growth. The fragmentation and the subsequent spheroidization of Si particles reduce their size and increase their number, leading to smaller and more uniformly distributed particles than in the as-cast condition. The variation of Si morphology causes an additional strengthening effect. LADOS et al [3] studied this phenomenon during T61 treatment of an Al−7Si−0.4Mg alloy in unmodified and Sr-modified condition. It is found that the fragmentation of Si particles and the consequent strengthening effect in the modified alloy take place in a shorter solution time than in the unmodified one. The modification of the melt with Sr is usually used, because it promotes the spheroidization of eutectic Si during the solidification of the alloy, thereby reducing the solution time needed to smooth the Si particles during the heat treatment. Heat treatment costs and energy consumption also decreased thanks to this process [1,2]. Although modified alloys are generally more prone to porosity, SHIVKUMAR et al [1] found that the tensile properties of the Sr-modified A356.2 alloy are higher than those of the unmodified alloy.

In Al−7Si−Mg alloys the high Si content also results in good fluidity, resistance to hot cracking and thermal expansion. The addition of Mg improves the specific strength and the yield strength, because it combines with Si to form the strengthening phases during the artificial aging.

Although Al−Si−Mg foundry alloys clearly differ in Si content from Al−Mg−Si wrought alloys, the strengthening mechanism is the same and so the precipitation sequence has usually been assumed to be similar. According to EDWARDS et al [4] the precipitation sequence is: Al SSS (supersaturated Al solid solution) → independent clusters of Si and Mg atoms → co-clusters of Si and Mg atoms→ small precipitates of unknown structure → β'' phase → β' + B' (Si-rich precipitate) phases → β phase. The β'' phase is considered the main strengthening phase in these alloys [4, 5].

Over the last 25 years, the heat treatment of Al−7Si−Mg alloys is of interest to both the research community and the industry, with several research papers [1−3,5−11] published on the topic. In recent years, various foundry processes have led to products with very thin parts, below the 4 mm threshold of the permanent mold casting technology. In order to obtain very thin parts with an almost traditional gravity casting process, some tweaks are needed to foundry practice and tools in order to fill the thin gaps in the mold. High pouring rates, specific surface coatings and a higher than usual die temperature enable parts as thin as 2 mm to be filled easily, but these conditions significantly change the solidification rate and thus the microstructure, the mechanical performances and the heat treatment response. Generally, this kind of casting is used as the crash and safety parts of a vehicle spaceframe, in order to reduce the overall mass of the car body and, consequently, tailored mechanical properties are needed to grant high elongation and strength at the same time. As the safety margins of these castings are reduced, the T6 heat treatment conditions adopted for the Al−7Si−Mg alloys have to be investigated to identify the best combination of strength and ductility. Furthermore, the cost and the production time associated with T6 heat treatment have to be optimized: shortening the total time of the heat treatment cycle and maintaining the performance of the components, have a considerable impact on productivity and manufacturing cost.

In the present work, the results of an experimental study carried out to optimize the solution treatment and artificial aging parameters in gravity cast thin bars of B356 aluminum alloy modified with Sr have been reported. These results, in terms of the best combination of mechanical properties and heat treatment duration, were correlated with the DSC analyses and were compared with those found in a previous investigation performed by the authors on the same alloy [7].

2 Experimental

The chemical composition of the Sr-modified B356 aluminum alloy is listed in Table 1.

Table 1 Chemical composition of alloy (mass fraction, %)

Si	Fe	Cu	Mn	Mg	Ti
7.43	0.078	0.008	0.010	0.331	0.106
B	Ca	Na	Sb	Sr	Al
0.0001	0.002	0.0008	0.001	0.016	Bal.

The alloy was provided as thin rectangular bars (100 mm × 24 mm × 3 mm) separately cast from automotive components by gravity casting. The hardness of the as-cast bars is HRF 61. The samples for DSC analyses were machined out of a bar prior to the heat treatments to prevent alterations due to the cut [12]. Their dimensions were approximately 4 mm × 4 mm × 1 mm and their average mass was 20−25 mg.

The bars and the DSC samples were heat-treated according to Table 2. Two solution temperatures were chosen: one just below and one just above 540 °C, a value recommended in most references (for example see Ref. [6]) and widely used in production for these alloys. The comparison with the results obtained by the authors in a previous work, where the solution temperature was 540 °C [7], is of interest in order to optimize the heat treatment conditions of the alloy. The artificial aging temperature was chosen from within the recommended range reported in [6]. The bars and the DSC samples were solutionized in a preheated electric oven, quenched in room temperature water and immediately refrigerated at −18 °C. The time lapse between quenching and refrigeration did not exceed 5 min to prevent detrimental effects on the mechanical properties of alloy due to natural aging [5,6]. Afterwards, the bars and the DSC samples were artificially aged in a preheated oven.

Table 2 Heat treatment conditions

Solution treatment		Artificial aging treatment	
Temperature/°C	Holding time/h	Temperature/°C	Holding time/h
530	4, 6, 8	165	6, 8, 16, 32
550	2, 3	165	2, 4, 6, 8, 16, 32

The influence of the solution heat treatment and artificial aging conditions on the mechanical properties of alloy was investigated through hardness and tensile tests. Dumb-bell specimens with a gauge length of 48 mm, a width of 12 mm and a thickness of 3 mm were extracted from the heat treated bars. The hardness tests were performed on the shoulders of the tensile specimens prior to the tests, following ASTM E 18-03 procedures [13]. A hardness tester Rockwell Rupac 500Mra, with a steel ball indenter diameter of 1.58 mm, a load of 588 N and a dwell time of 15 s was used. Ten measurements were taken and averaged for each heat treatment condition. The tensile tests were carried out on three samples for each heat treatment condition, following UNI EN ISO 6892-1:2009 [14]. An electromechanical testing machine Instron 3369 at a strain rate of 0.4 mm/min was used. The best combination of mechanical properties and T6 heat treatment duration was identified. For metallographic observations, an as-cast bar and a tensile sample heat treated in the best condition were sectioned orthogonally to their axis and were prepared with standard metallographic techniques (ground with SiC papers and polished with 1 μm diamond paste). Image analysis of eutectic silicon fraction was performed using LAS 4.0 software integrated with the optical microscope. The analysis was carried out on samples treated with the lower, the average and the higher aging time for each solution treatment condition, selected in order to cover the whole range of parameters tested. For each condition, at least 5 images with a 200× magnification were analyzed. The various heat treatment parameters influence the content and distribution of the alloying elements in solid solution, also they can control the microstructure and hence the mechanical properties. The spheroidization and coarsening of the Si particles are some of the observed changes. In this context, it is convenient to have a parameter for describing these changes, being the equivalent diameter and the shape factor among the most used. For this purpose the roundness was the parameter considered to evaluate the influence of different solution treatments on the microstructure, indeed this shape factor's parameter, is equal to 1 when the object is a perfect circle and gives important indications about the spheroidization of the particles. The microstructure was examined using a Leica DMI 5000M optical microscope. Finally, the effect of the heat treatment conditions on the precipitation sequence was investigated by DSC. The analyses were carried out on the samples, heat treated at the best solution temperature. They were performed using a TA Instrument DSC Q100 in a temperature range from −50 °C to 350 °C, with a heating rate of 10 °C/min and operating in nitrogen atmosphere. The thermal events associated with the phase transformations were analyzed after the subtraction from the experimental results of the baseline run performed with empty sample holders.

3 Results and discussion

3.1 Hardness and tensile properties

The hardness of the Sr-modified B356 specimens solubilized at 530 °C and 550 °C for various solution time is shown as a function of aging time in Figs. 1(a) and (b), respectively. The hardness slightly decreases (approximately 5 HRF) when the solution time increases from 2h to 3h for 550 °C solution temperature, whereas it does not show any significant dependence on solution time for 530 °C solution temperature. The hardness also slightly increases with the solution temperature. A similar effect of solution time on the hardness of A356 was also observed by SHIVKUMAR et al [15]. Furthermore, it produces an earlier peak hardness at a higher solution temperature compared with that at a lower solution temperature: the hardness peak of the 550 °C solution temperature is reached after 8 h of aging, whereas 16 h of aging are required for solutionizing at 530 °C.

The main requirements for temperature and duration of the solution treatment are that the former should be sufficiently high and the latter sufficiently long to completely dissolve the equilibrium β phase formed during casting and to allow diffusion of alloy elements from the dissolved phase out into the Al matrix. The best solution treatment conditions give the highest hardness and strength after aging treatment. The curves of Fig. 1 show that the shortest solution time (4 h and 2 h for 530 °C and 550 °C solution temperature, respectively) seems to be enough to obtain both the described phenomena. A prolonged solution treatment does not lead to increased hardness.

As expected, the hardness of alloy is strongly influenced by the aging time: it increases very rapidly with increasing time at the beginning of aging, it reaches a maximum value and then slightly decreases only for the lowest solution time at the lowest solution temperature. The increase of hardness is due to the progressive formation of coherent precipitates (β'' phase is the main strengthening phase), whereas its decrease is due to the subsequent formation of non-coherent precipitates (the equilibrium β phase). A similar maximum hardness was obtained with 4 h solutionizing at 530 °C, followed by a room temperature water quench and then 16 h artificial aging at 165 °C (HRF 86); or with 2 h solutionizing at 550 °C, followed by a room temperature water quench and then 8 h artificial aging at 165 °C (HRF 89). These results can be compared with those obtained in our previous work [7], where the same alloy was solutionized at 540 °C for 6 h, water quenched at room-temperature and then aged at 155, 165 and 180 °C in a range of 40 min to 32 h. Similar maximum hardness (approximately HRF 88) was obtained for all aging temperatures, the hardness peak at 165 °C aging temperature was reached after 8 h.

The tensile test results of the specimens solubilized at 530 °C and 550 °C for various solution times are shown as a function of aging time in Figs. 2(a) and (b), respectively. The yield strength ($R_{p0,2}$) and the ultimate tensile strength (R_m) of the specimens solubilized at 530 °C slightly improve with solution time, although hardness does not depend on it. The maximum $R_{p0,2}$ increases from 220 to 240 MPa and the maximum R_m from 270 to 283 MPa by increasing the solution time from 4 to 8 h. Therefore, it can be concluded that the shortest solution time at 530 °C is not enough to complete the dissolution and homogenization phenomena and the optimal solution time is 8 h. Increasing the solution temperature to 550 °C accelerates both processes, so it is sufficient for the alloy to achieve the optimum strength in 2 h at 550 °C ($R_{p0,2}$ ~240 MPa and R_m ~290 MPa). Figure 2 also shows that the yield strength and the ultimate tensile strength improve with aging time, attain a maximum and in some cases slightly decrease. The trend of yield strength as a function of aging time is similar to that of hardness because both properties depend on the progressive precipitation of the strengthening phases with increasing the aging time.

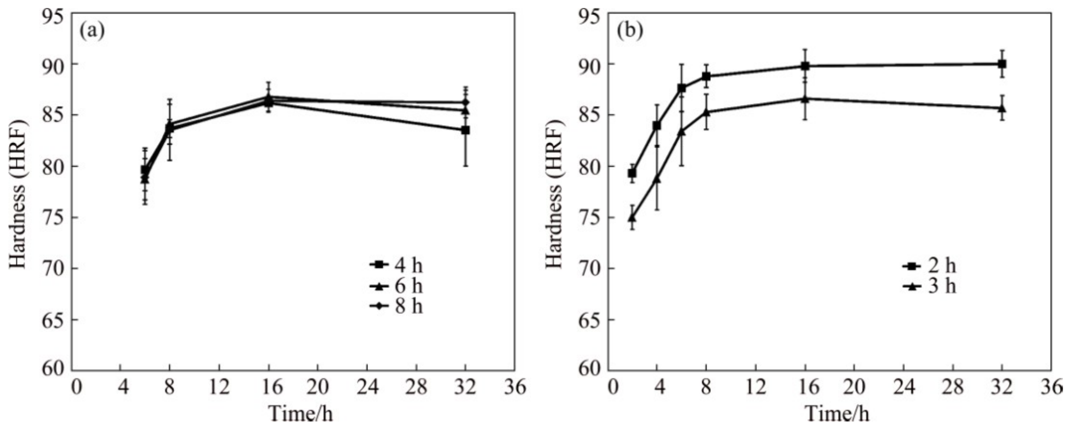

Fig. 1 Hardness of Sr-modified B356 alloy solubilized at 530 °C (a) and 550 °C (b) for various solution time as function of aging time (aging temperature 165 °C)

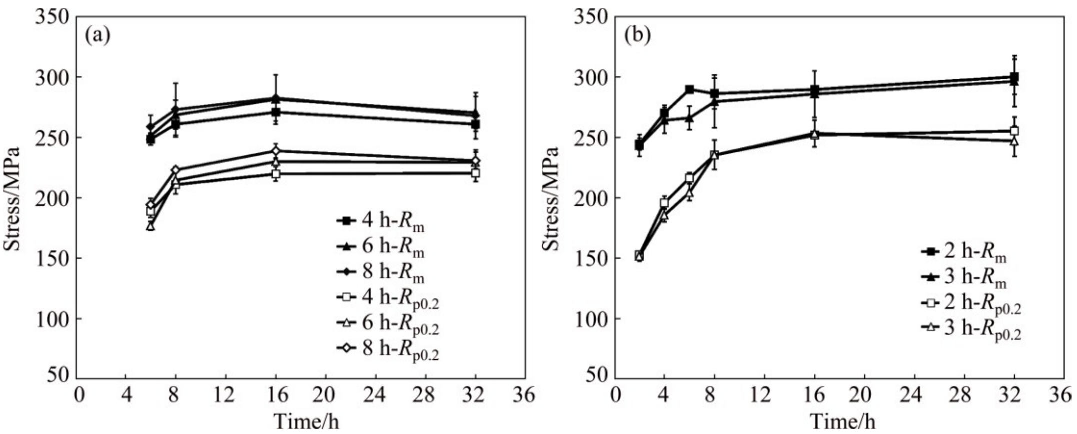

Fig. 2 Tensile strength (R_m) and yield strength ($R_{p0.2}$) of Sr-modified B356 alloy solubilized at 530 °C (a) and 550 °C (b) for various solution time as function of aging time (aging temperature 165 °C)

The maximum hardness and strength are reached when β'' phase is precipitated.

The best combination of $R_{p0,2}$ and R_m can be obtained in two ways. The first is with a solution treatment carried out at 530 °C for 8 h, followed by quenching and artificial aging at 165 °C for 16 h. The second is with a solution treatment carried out at 550 °C for 2 h, followed by quenching and artificial aging at 165 °C for 8 h. In our previous work [7] similar maximum values of $R_{p0,2}$ and R_m were found after all heat treatments at any aging temperature investigated. The peak of both properties at aging temperature of 165 °C ($R_{p0,2}$=240 MPa and R_m=290 MPa) was reached after 8 h (the solution temperature was 540 °C and the solution time was 6 h). It can also be observed that $R_{p0,2}$ and R_m slightly improve by increasing the solution temperature from 530 to 540 °C at the same aging time, the solution time being equal to 6 h. This result is consistent with that found by SHIVKUMAR et al [1]. The authors studied the effects of solution temperature on the tensile properties of a Sr-modified A356.2 alloy cast in a permanent mold. The solution temperatures investigated were 540, 550, 560 °C and the solution time was 100 min. The solution treatment was followed by quenching in water at 60 °C, natural aging at room temperature for 24 h and artificial aging at 171 °C for 4 h. The authors found that the tensile properties slightly increase when increasing the solution temperature from 540 to 550 °C, whereas they decrease at the solution temperature of 560 °C and above 560 °C the first liquid begins to form at the grain boundaries.

Reduction of solution treatment and aging time in these alloys means reducing heat treatment costs and energy consumption. Therefore, the long duration required to obtain the maximum hardness and strength with a solution treatment carried out at 530 °C makes this solution temperature not optimal. Finally, a solution treatment of 2 h at 550 °C, followed by water quenching and artificial aging at 165 °C for 8 h is enough to achieve the best mechanical properties in the Sr-modified B356 thin bars that we investigated.

3.2 Metallographic observations and DSC analysis

Figure 3(a) shows the microstructure of Sr-modified B356 alloy in the as-cast condition: it consists of primary α(Al) dendrites and interdendritic Al−Si eutectic regions. Some iron-rich particles are also observed under high magnification, as highlighted in Fig. 3(b). The modified Si particles appear small and round in 2D images, but, actually they are fibrous and branched and have a coral-like structure in 3D, as observed by LADOS et al [3] in an Al−7Si−0.4Mg alloy modified with 0.019% Sr.

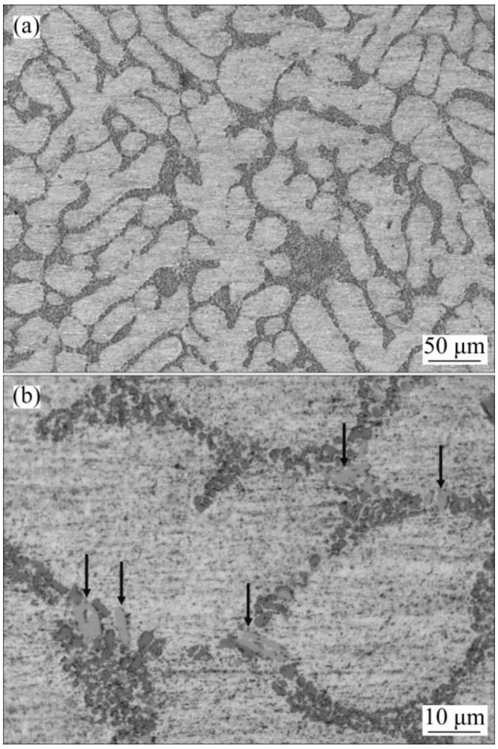

Fig. 3 Microstructure of Sr-modified B356 alloy in as-cast condition at low magnification (a) and high magnification (b) with iron-rich particles highlighted by arrows

Figures 4 and 5 show the alloy microstructure, at different magnifications, after the solution treatment carried out at 550 °C for 2 and 3 h followed by quenching and 165 °C artificial aging for 2, 8 and 32 h and after the solution treatment carried out at 530°C for 4, 6 and 8 h, followed by quenching and 165 °C artificial aging for 6, 16 and 32 h.

Figure 6 shows the alloy microstructure after the heat treatment which gives the best mechanical properties (solution treatment carried out at 550 °C for 2 h, followed by quenching and 165 °C artificial aging for 8 h), at low and high magnification.

The typical globular shape of the Si particles can be observed from Figs. 4, 5 and 6. The modification with Sr ensured a fast fragmentation and spheroidization of the Si particles and meant that a shorter solution treatment could be used. The fragmentation and the spheroidization of the Si particles reduce their size and increase their number, leading to smaller and more uniformly distributed particles than in the as-cast condition. The variation of Si morphology causes an additional strengthening effect, as described in Ref. [3]. PAN et al [16] showed that in A357 alloys the maximum yield strength and ultimate tensile strength were reached at time corresponding to the Si morphology (in two dimensional images) attained at the late stage of spheroidization or early stage of coarsening. The iron-rich intermetallic particles (Fig. 6(b)) are rounder

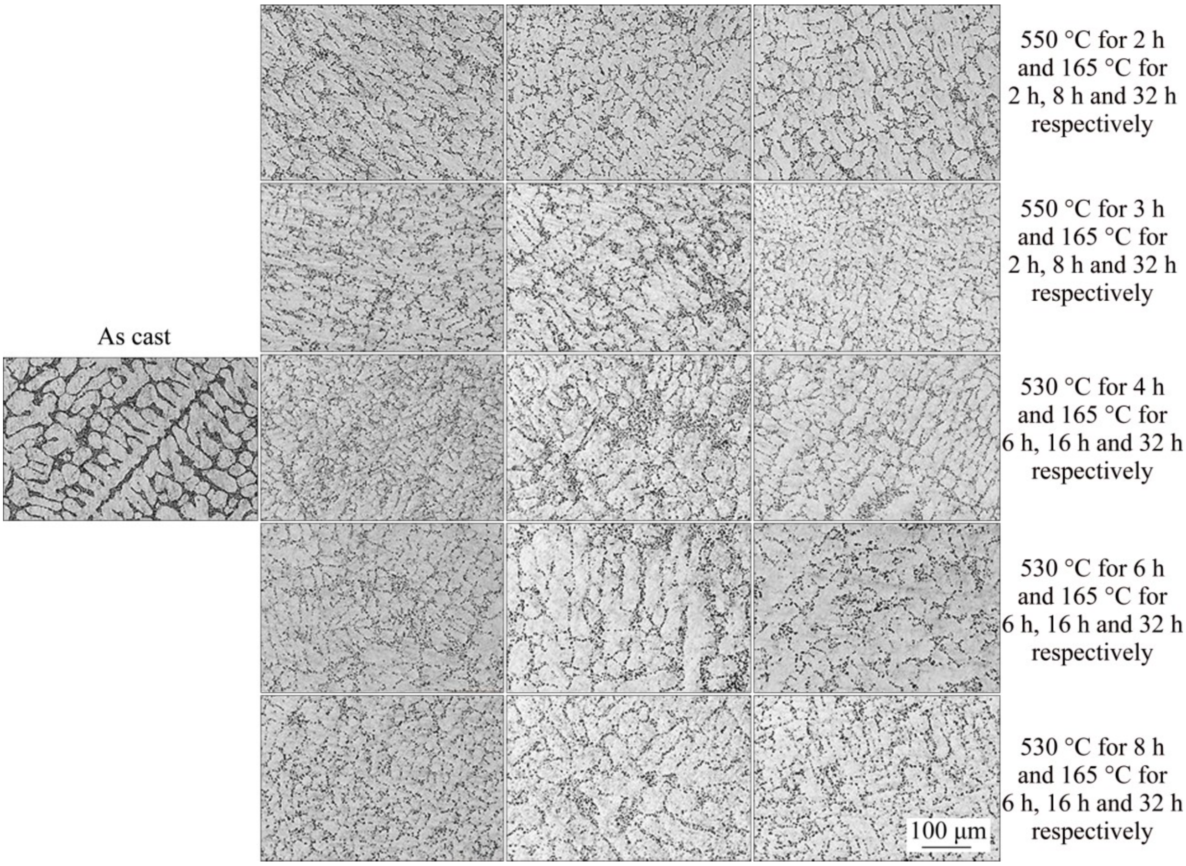

Fig. 4 Some examples of microstructures of B356 aluminum alloy obtained with different heat treatment parameters

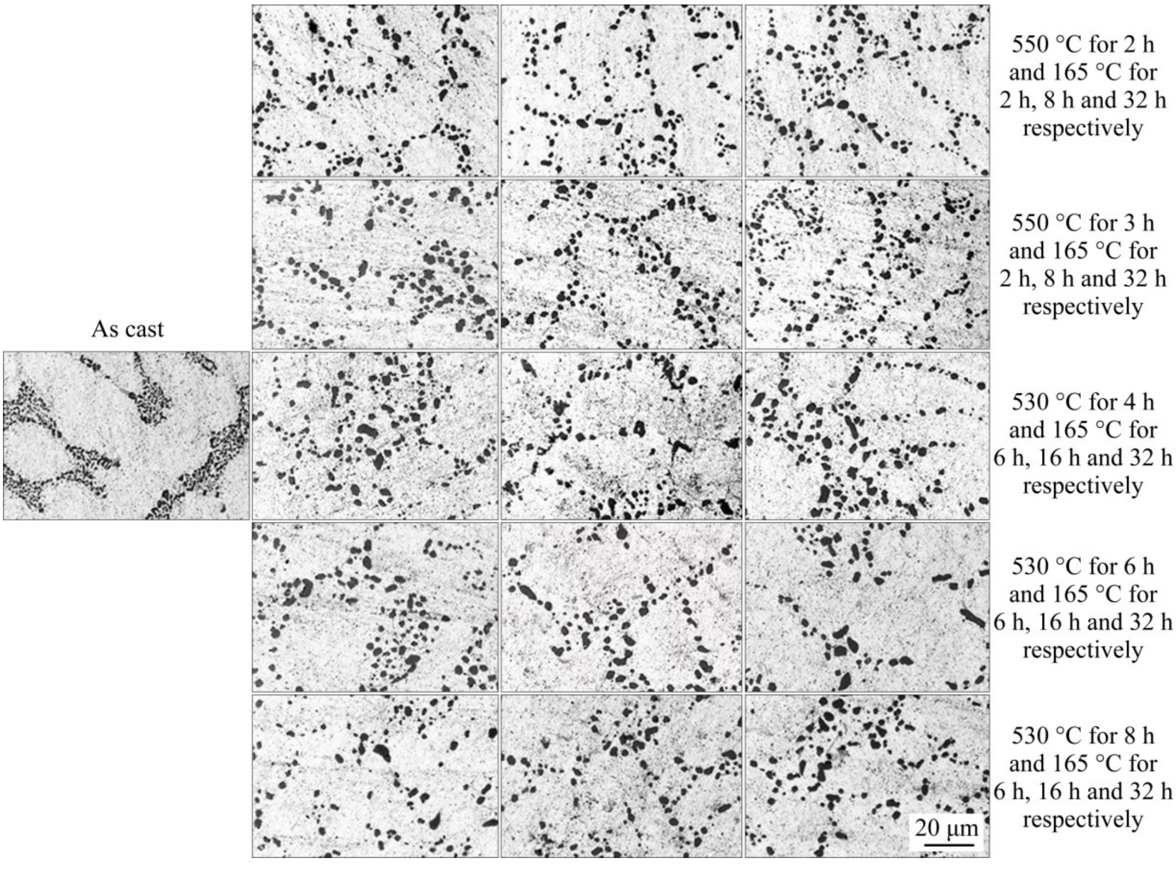

Fig. 5 Some examples of microstructures of B356 aluminum alloy obtained with different heat treatment parameters

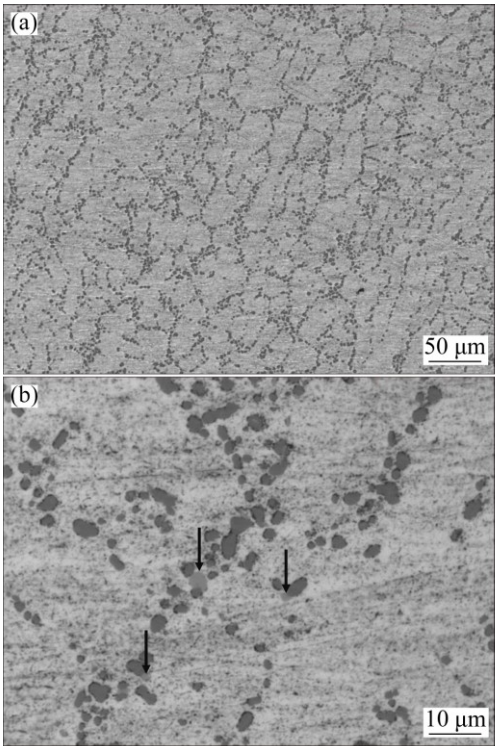

Fig. 6 Microstructure of Sr-modified B356 alloy after 550 °C solution treatment for 2 h, followed by quenching and 165 °C artificial aging for 8 h (At low magnification (a) and at high magnification (b) with iron-rich particles highlighted by arrows)

and smaller than those observed in the as-cast condition (Fig. 3(b)).

The results of the image analysis on the samples heat treated in the selected conditions are shown in Fig. 7. In particular, the roundness of the eutectic Si related to the aging time was evaluated. The spatial distribution and shape of the Si particles is slightly modified, due to the solution treatment on the already finely dispersed and Sr modified solidification microstructure. Si shape and distribution image analysis results show that an increasing Si particle coalescence and smoothing happens as a function of increasing solution treatment time while almost no influence is exerted by the following aging phase. The results of the roundness factor confirmed the output of mechanical tests; in fact the best response was obtained with the same combination of heat treatment parameters ((530 °C, 8 h) + (165 °C, 16 h) and (550 °C, 2 h) + (165 °C, 8 h)) that gave the best R_{p02} and R_m.

Figure 8 shows the DSC curves of the Sr-modified B356 alloy solubilized at 550 °C for 2 and 3 h, with artificial aging carried out at 165 °C for various time. The curves were shifted along the y-axis to help the observation. Peak identification of Al−Si−Mg alloys DSC thermograms involves many techniques and is not

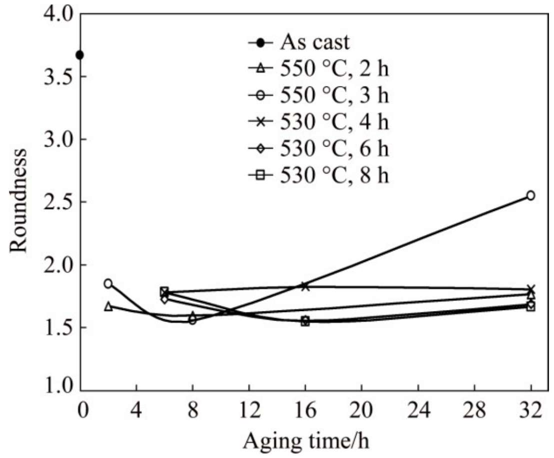

Fig. 7 Roundness as function of aging time for Sr-modified B356 alloy in as-cast condition after different solution treatment conditions

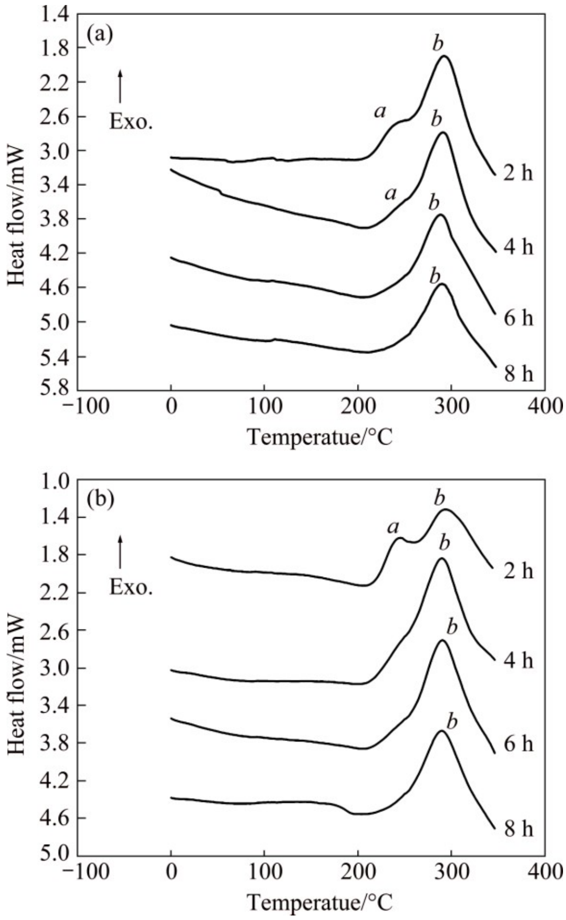

Fig. 8 DSC curves of Sr-modified B356 alloy solubilized at 550 °C for 2 h (a) and 3 h (b), with aging at 165 °C from 2 to 8 h

unique among the papers in the current literature. Therefore, it will be performed with the support of previous works [4,7,17−19], in some cases carried out on 6xxx series alloys, focusing on the most relevant information linked to the alloy B356.

Two distinct exothermic peaks can be observed: around 250 °C (peak *a*) and around 300 °C (peak *b*). The exothermic peak *a* was observed only in some DSC curves, implying that it was caused by the formation of the β'' phase, according to the results of [7,17]. The height of this peak decreases with increasing aging time and the peak completely disappears after 4 h, for both solution time considered. This trend suggests that the β'' phase is fully formed after 4 h of artificial aging, for both solution time. This conclusion can be drawn thanks to previous works, such as that of ESMAEILI et al [17], where DSC analysis was supported by quantitative TEM.

The exothermic peak *b* appears to be caused by the transition from β'' phase to β' and B' (Si-rich precipitate) phases, as reported by CESCHINI et al [5]. This peak does not seem to be affected by artificial aging.

These results are consistent with those obtained by the hardness and tensile tests, the reduction of peak *a* in DSC curves, as well as the increase in hardness (Fig. 1(b)) and in strength (Fig. 2(b)), with aging time can be associated with the gradual precipitation of the β'' phase during the artificial aging. These effects confirm that β'' is the main strengthening phase in B356 alloy. In particular, the maximum mechanical properties were obtained when β'' is completely precipitated during aging.

The expected decrease in hardness and strength caused by the formation of the equilibrium β phase was not observed in the aging time range investigated in this study. The DSC analysis carried out on the samples solubilized at 550 °C confirmed that the equilibrium β phase did not precipitate at any artificial aging time evaluated.

4 Conclusions

1) The hardness of the bars slightly decreased (approximately HRF 5) when the solution time increased from 2 to 3 h at the solution temperature of 550 °C, whereas it did not show any significant dependence on solution time for the 530 °C solution temperature,

2) The tensile strength of the specimens solutionized at 550 °C did not show any significant dependence on the solution time, whereas the strength of the specimens solutionized at 530 °C slightly increased with it.

3) Both hardness and tensile strength slightly increased with the increase of the solution temperature.

4) Both properties improved with artificial aging time to a maximum and then in some cases slightly decreased.

5) The best combination of mechanical properties and heat treatment duration was obtained with 2 h solutionizing at 550 °C and 8 h aging at 165 °C.

6) The influence of the heat treatment conditions on the precipitation sequence of the alloy was investigated by DSC. The mechanical properties of alloy were correlated with the DSC results and it was found that the maximum mechanical properties were obtained when β'' was completely formed during aging.

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重力铸造 Sr 变质 B356 铝合金的热处理工艺优化

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摘　要：近年来，采用某些铸造工艺使生产薄壁件成为可能，尺寸上已实现金属型铸造技术能实现小于 4 mm 的极限。因为这些铸件的安全边际量减小，有必要研究 Al−7Si−Mg 合金的 T6 热处理条件，以获得最佳的强度和延性组合。此外，必须优化与 T6 处理有关的成本和生产时间。采用实验研究优化了重力铸造 Sr 变质 B356 铝合金薄壁件的固溶强化和人工时效条件。分别选取 530 °C 和 550 °C 作为固溶处理温度，固溶 2~8 h，随后水淬，165 °C 人工时效 2~32 h。获得最佳的综合力学性能的热处理条件是：550 °C 固溶 2 h，165 °C 人工时效 28 h。DSC 分析表明，β''相在人工时效过程中充分析出后，合金达到最佳力学性能。

关键词：B356 铝合金；重力铸造；热处理；硬度；拉伸性能；β''相

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