Effect of magnesium content on the ageing behaviour of water-chilled Al–Si–Cu–Mg–Fe–Mn (380) alloy castings

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A study of the effect of magnesium concentration on the ageing behaviour as measured by the hardness of 380 alloy was conducted for three levels of magnesium, namely 0.06 (base alloy), 0.33 and 0.5 wt%, for water-chilled castings (dendrite arm spacing $\sim$ 10–15 μm). Differential scanning calorimetry analysis of as-cast samples was carried out to determine the changes in the reactions of the phases obtained during alloy solidification, employing heating rates of 0.1 and 1.0 °C s^{-1}, up to approximately 700 °C. Two heat treatments were applied to the as-cast alloys: T5 comprising ageing at 25 (room temperature), 155, 180, 200 and 220 °C, for times up to 200 h, and T6 comprising solution heat treatment at 480 °C or 515 °C for 8 h, followed by quenching in warm water at 60 °C, followed by immediate artificial ageing at 155 or 180 °C for varying times up to 100 h. The results show that the higher hardness values obtained with T6 treatment can be explained by the excess precipitation of magnesium-containing phases in the as-solidified alloys. This precipitation could be eliminated under the high cooling-rate conditions prevalent in die-casting operations so that T5 treatment may be used to replace T6 treatment to produce the same hardness values. In addition, solution heat treatment in the low-temperature range (480–515 °C) is adequate to produce the required changes in silicon morphology and dissolution of magnesium in the matrix. No significant difference in hardness behaviour was observed when the magnesium content was increased beyond 0.3 wt%.

1. Introduction

One of the most widely used aluminium die casting alloys in the automotive industry is the 380-type alloy, belonging to the Al–Si–Cu family. Many factors affect the mechanical properties of these alloys, including the alloying elements and their concentrations, and the applied heat treatment [1–3]. Magnesium, in particular, aids in enhancing the strength of these alloys, but at the expense of ductility. With a higher magnesium content in standard 380 alloys (i.e. above 0.1 wt %), and with the application of a suitable heat treatment, alloys with higher strength and hardness can be produced, having better wear resistance [4]. The heat treatment can involve either a low-temperature age-hardening (T5-type) or a solution treatment + age-hardening (T6-type) process, that are particularly suitable for alloys containing 0.4%–0.6 % Mg, the underlying mechanism being the formation of Mg_2Si, a strong hardening constituent. However, in the case of 380-type alloys, increasing the magnesium content must be considered in careful conjunction with the copper content of the alloy, as copper is also a strong hardening constituent [4]. The influence of the alloying elements on the structural development and hence on the properties can be elucidated from an examination of the resulting microstructures [5] and by the popular medium of thermal analysis, whereby the solidification process of the alloy can be monitored [6].

The present study was carried out to determine the effect of magnesium concentration on the ageing behaviour, as measured by the hardness, of 380 alloy at three levels of magnesium, 0.06 (base alloy), 0.33 and 0.5 wt %, using water-chilled castings (dendrite arm spacing $\sim$10–15 μm). The results are presented and discussed.

2. Experimental procedure

The base 380 alloy was supplied by Alcan Ingot Alloys Canada, Guelph, Ontario, with the chemical composition shown in Table I. The alloy was supplied in the unmodified form. About 6 kg of the material were melted at a time, and to each melt, measured quantities of magnesium were added in the form of Al–50 wt % Mg master alloy to obtain final magnesium levels of 0.33 and 0.5 wt %. Each melt was degassed at 740 °C for 20 min by passing dry argon through a lance. The hydrogen level of the melt after degassing was found to be about 0.10 ml/100 g Al, using an Alscan^{TM} hydrogen analyser. No melt treatment such as modification or grain refinement was

TABLE I Chemical composition of the as-received 380 base alloy

	Si	Cu	Mg	Fe	Mn	Ti	Zn	Ni	Cr	Al
Specified conc. (wt %)	7.50–9.50	3.00–4.00	0.10	1.00–1.30	0.50	–	3.00	–	–	Bal.
Chemical Analysis (wt%)	9.18	3.22	0.06	1.01	0.16	0.02	2.28	0.04	0.03	Bal.

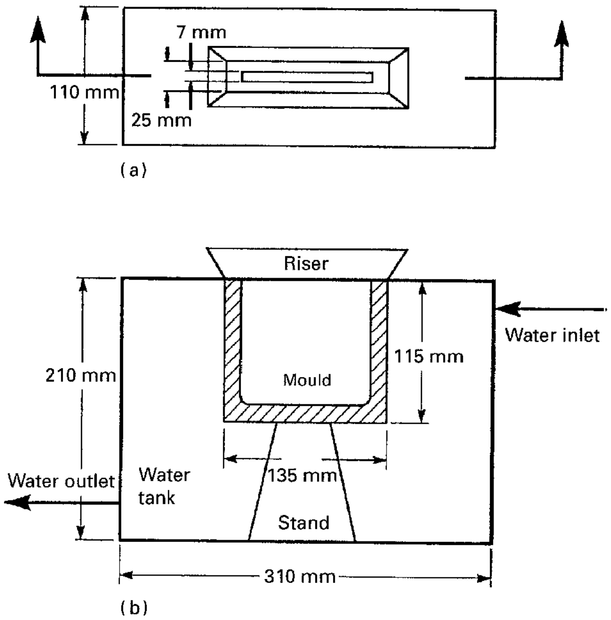


Figure 1 Schematic diagram of the set-up used for the water-chilled castings. (a) Top view, (b) section view.

used as these treatments are not generally required in die casting operations where the high cooling rate itself provides the necessary structural refinement.

The degassed material was poured into a graphite mould of 5 cm diameter, maintained at 400 °C (taken as the reference), and a rectangular copper mould placed in a tank of running cold water (a schematic diagram of the apparatus is shown in Fig. 1). This arrangement permitted two different cooling conditions to be obtained.

Powders for differential scanning calorimetric (DSC) measurements were prepared from the as-solidified materials (referred to hereafter as 380 (base alloy), 380 + 0.33 Mg, and 380 + 0.5 Mg, and corresponding to 380 alloy containing 0.06, 0.33 and 0.5 wt % Mg, respectively) and examined in a DuPont 2100 apparatus using different heating rates (0.1 and 1 °C s^{-1}). DSC runs were initiated at room temperature and completed at approximately 700 °C. The output was in mW, and the net heat flow to the reference (high-purity annealed aluminium) relative to the sample was recorded as a function of temperature, with the base line subtracted from the data.

Heat treatments were performed in an air-forced oven (± 1 °C). Samples were cut from the water-chilled cast blocks normal to the solidification direction and heated at 0.5 °C s^{-1} up to the quenching temperature. Two solution temperatures were used, 480 and 515 °C, keeping the solution time constant at 8 h, followed by quenching in hot water at 60 °C. As-cast as well as solution-treated samples were naturally and artificially aged at 25, 155, 180, 200 and 220 °C for varying times up to 200 h. The same oven was employed for the artificial ageing.

Microstructural details of prepared samples (sample size 7 mm × 10 mm × 60 mm, properly polished to obtain a smooth surface) were analysed using an Olympus PMG3 optical microscope. Hardness measurements were made on both sides of each sample using a Brinell hardness tester (500 kg charge and 10 mm diameter steel balls). An average of at least six readings were taken for each recorded measurement.

3. Results and discussion

3.1. Solidification and remelting characteristics

Fig. 2 shows the radiograph of a plate sample taken from a 380 ± 0.5 Mg alloy casting. The plate contains a homogeneous area towards the centre of the casting as indicated by the uniform grey matt appearance. The circular flow pattern on the print is probably caused by the turbulent flow of the metal into the cold mould (~7 °C) during casting.

From published results [6–8], the expected sequence of reactions taking place for the 380 base alloy at slow cooling rates (~ 0.2 °C s^{-1}) is:

1. development of dendritic network (α-aluminium): 578–577 °C;
2. precipitation of β-Al_5FeSi iron phase: 572–564 °C;
3. main eutectic reaction involving silicon- and α-$Al_{15}(Fe, Mn)_3Si_2$ iron phase: 561–559 °C;
4. precipitation of Mg_2Si: 559–491 °C;
5. precipitation of Al_2Cu: 491–487 °C;
6. formation of complex eutectic containing Al_2Cu and $Al_5Mg_8Si_6Cu_2$: 485–479 °C.

The cooling curves obtained for the 380, 380 + 0.33 Mg, and 380 + 0.5 Mg alloys are shown in Fig. 3a and b for graphite and metallic moulds, respectively. The curves fail to show the peak corresponding to the Mg_2Si phase. The results show that increase in the magnesium content up to 0.3 wt % does not affect the solidification behaviour. The eutectic temperature changed by approximately 1 °C for each 0.1 wt % Mg added. The copper phase temperature (Reaction 5) is observed to be shifted to a lower temperature. The effect of magnesium content on the eutectic and post-eutectic reactions is listed in Table II, while Table III shows the solidification characteristics of the 380 alloys with different magnesium concentrations.

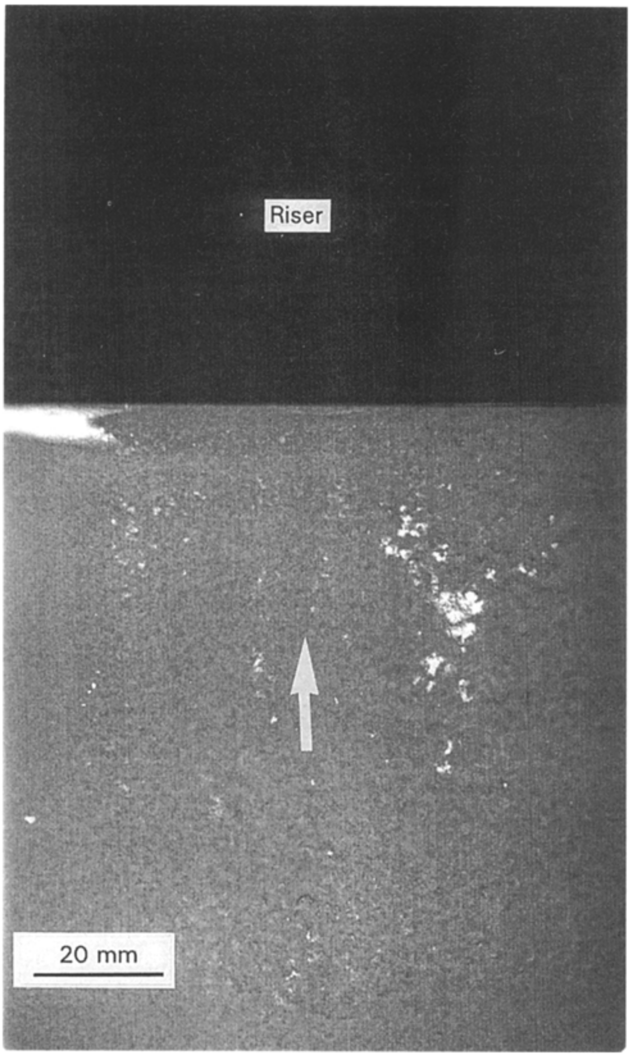


Figure 2 Radiograph of 380 + 0.5Mg alloy, arrow indicates the solidification direction.

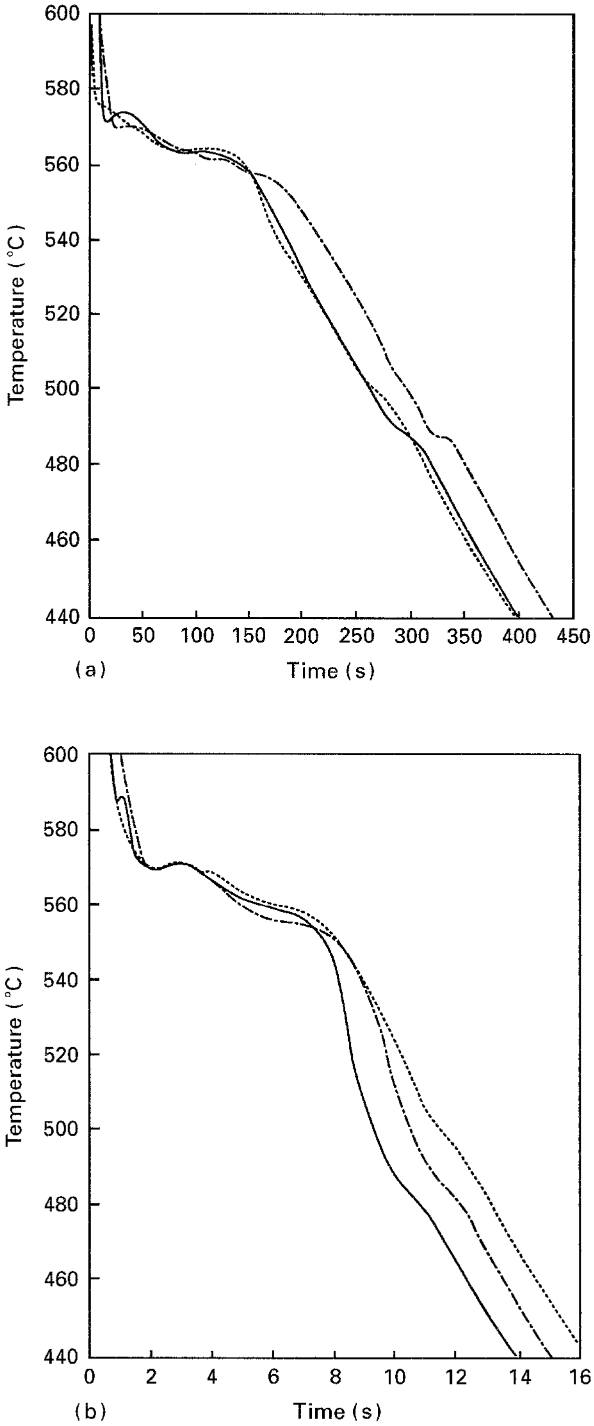


Figure 3 Cooling curves obtained for the three alloys, (——) 0.33, (–·–) 0.5 and (– – –) base 0.06 wt % Mg, at two different cooling rates: (a) 0.4 °C^{-1}s; (b) 10 °C s^{-1}.

Fig. 4a–c are typical differential scanning calorimetry (DSC) plots obtained for the three 380 alloys (containing 0.06, 0.33 and 0.5 wt % Mg), respectively. These alloys were solidified at 0.4 °C s^{-1} and reheated at 0.1 °C s^{-1}. Table IV indicates the temperature of formation of various precipitates. The data reveal that both the Mg_2Si and Al_2Cu + $Al_5Mg_8Si_6Cu_2$ reactions are shifted to lower temperatures when the magnesium level is increased to 0.33 wt %, with an increase in the melting temperature range. The latter is caused by the depression in the start-of-melting temperature, l_s, and increase in the end-of-melting temperature, l_f. Further increase in magnesium content, i.e. up to 0.5 wt %, does not seem to have a significant influence on the melting characteristics of this alloy.

Fig. 5a–c are the DSC plots produced from the same samples at a heating rate of 10 °C s^{-1}. The main observations to be noted from these plots are:

(a) all reactions are shifted to higher temperatures, contrary to the solidification behaviour observed/mentioned above, and which could result from (b);

(b) both Al_2Cu and Mg_2Si reactions are joined in a continuous peak (in some cases, the Mg_2Si peak

TABLE II Effect of magnesium content on the eutectic and post-eutectic reactions [7]

Cooling rate (°C s^{-1})	Mg (wt %)	Eutectic reaction temperature (°C)	Post-eutectic reaction temperature (°C)
~ 0.4	0.06	564.4	523, 498.7, 464
	0.33	563.6	492.6
	0.50	561.7	488
~ 10	0.06	560.0	496.8
	0.33	560.3	487.0
	0.50	556.5	484.0

TABLE III Solidification characteristics of the 380 alloys with different magnesium concentrations (wt %)

Solidification parameters	Cooling rate $\sim 0.4\,°C\,s^{-1}$			Cooling rate $\sim 10\,°C\,s^{-1}$		
	0.06Mg	0.33Mg	0.5Mg	0.06Mg	0.33Mg	0.5Mg
Start of solidification temperature (°C)	572.5	570.0	567.5	568.7	568.0	568.0
Solidification temperature range (°C)	73.8	77.4	79.5	71.9	81.0	81.0
Solidification time (s)	263.0	283.3	310.0	10.5	10.5	9.2
End of solidification temperature (°C)	498.7	492.6	488.0	496.8	487.0	487.0

TABLE IV Remelting characteristics of the three alloys used, solidified at $0.4\,°C\,s^{-1}$

Mg (wt %)	Heating rate ($°C\,s^{-1}$)	t_s^a (°C)	Al_2Cu^b (°C)	Mg_2Si (°C)	Si_{eut} (°C)	β-iron (°C)	t_f^a (°C)	Melting range (°C)
0.06	0.1	508.2[c]	511.8[c]	527.23[c]	568.84	576.5	588.2	80.4
0.33		490.5	500.8	516.87	566.45	576.5	590.8	100.3
0.50		492.9	500.45	515.18	564.47	576.0	591.0	98.1
0.06	1.0	520.0	526.56	537.0	577.53	N/A	608.2	88.2
0.33		502.3	515.94	528.57	577.81	N/A	612.9	110.6
0.50		502.3	516.81	527.6	575.95	N/A	611.76	109.6

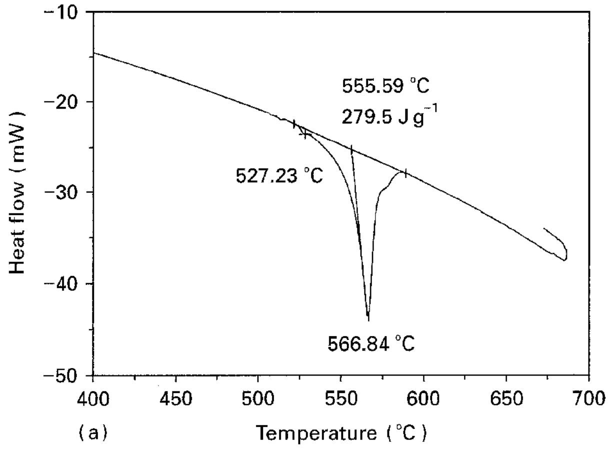
[a] Defined by the deviation of the curve from the base line.
[b] With or without $Al_5Mg_8Si_6Cu_2$.
[c] Identification not certain.

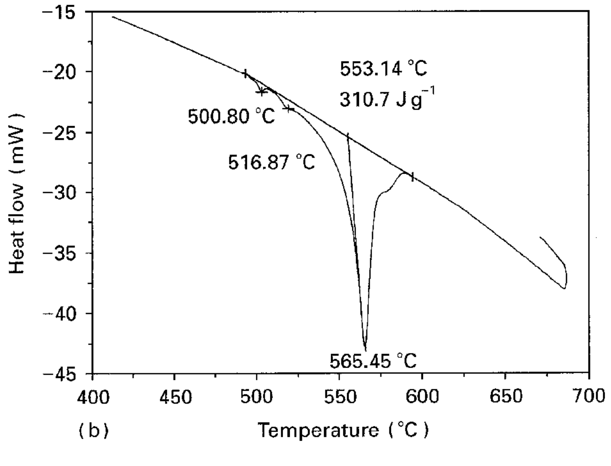
cannot be distinguished clearly or appears in the form of a hump); and

(c) an increase in the melting temperature range.

The heating curves ($0.1\,°C\,s^{-1}$) of samples solidified at $10\,°C\,s^{-1}$ reveal the presence of a single peak at low temperatures, Fig. 6a–c. The temperature range (Table V) is very close to that obtained for Mg_2Si, for samples solidified at $0.4\,°C\,s^{-1}$. Thus, increasing the solidification rate may possibly result in eliminating the Mg_2Si reaction, with the excess magnesium being precipitated as $Al_5Mg_8Si_6Cu_2$ (Reaction 6) instead, at the end of solidification. This observation was confirmed by microstructural examination, as will be discussed in the next section.

The heat flow–temperature plots for samples solidified at $10\,°C\,s^{-1}$, and using a heating rate of $1\,°C\,s^{-1}$ (Fig. 7a–c) are found to be more or less the same as those obtained for samples solidified at $0.4\,°C\,s^{-1}$, and heated at the same rate. However, there is a marginal increase in the melting temperature range ($\sim 10\,°C$), Table VI.





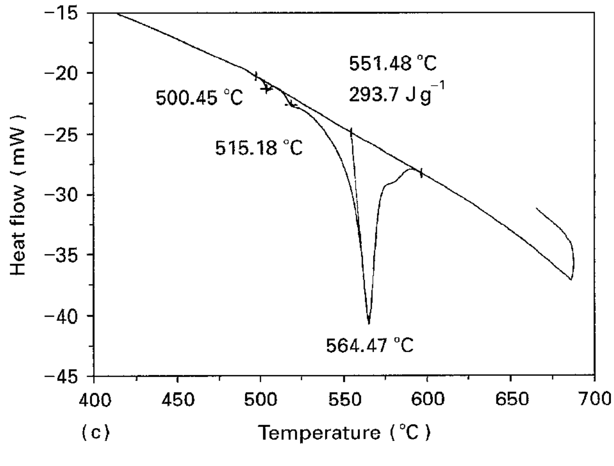


Figure 4 DSC curves of samples solidified at $0.4\,°C\,s^{-1}$ and reheated at $0.1\,°C\,s^{-1}$: (a) 0.06, (b) 0.33, (c) 0.5 wt % Mg.

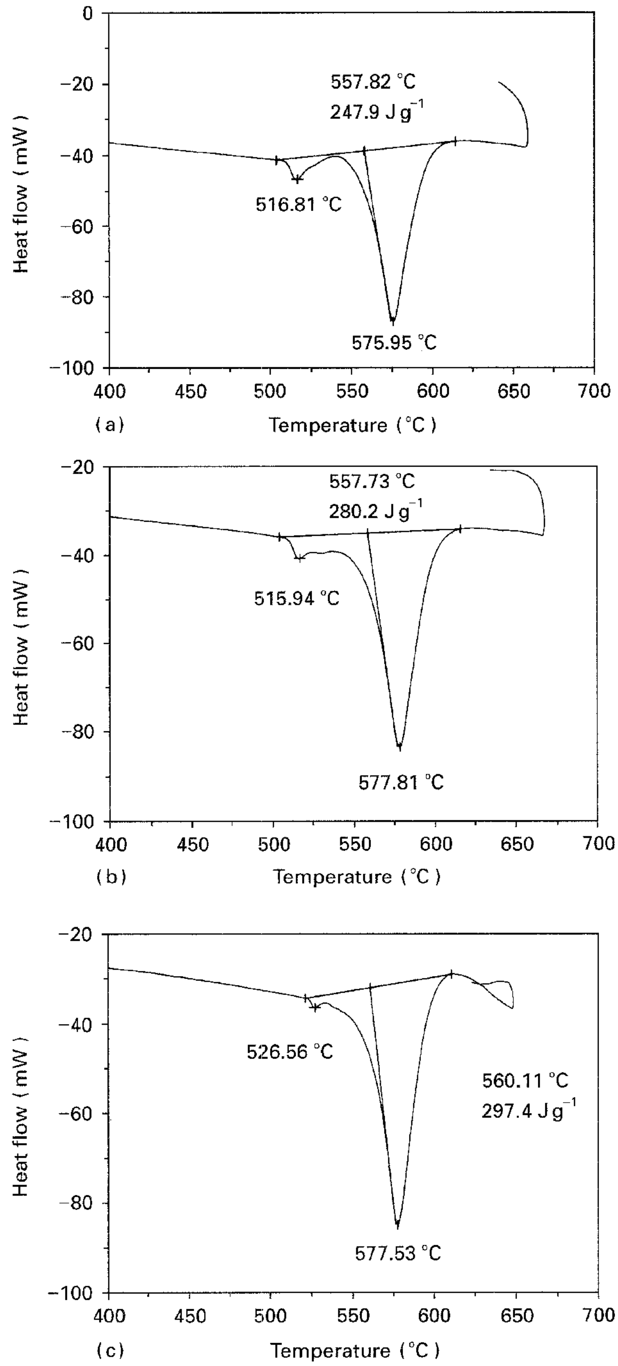


Figure 5 DSC curves of samples solidified at 0.4 °C s^{-1} and reheated at 1 °C s^{-1}: (a) 0.06, (b) 0.33, (c) 0.5 wt % Mg.

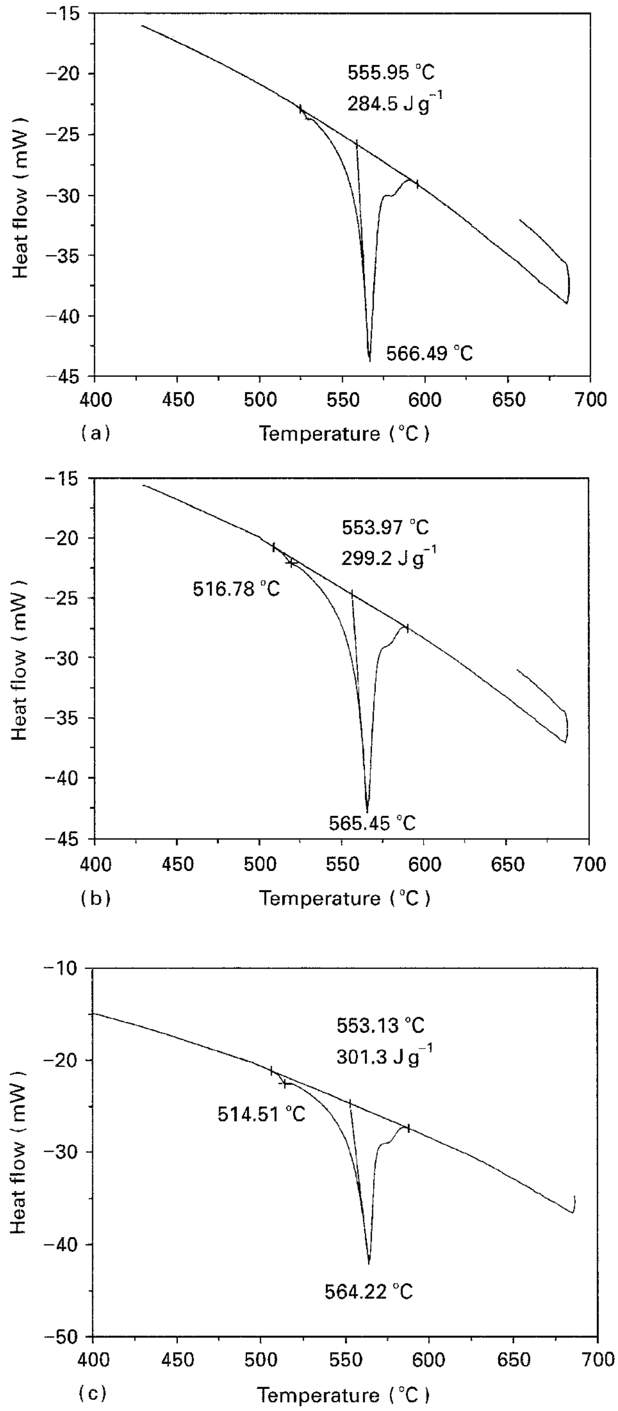


Figure 6 DSC curves of samples solidified at 10 °C s^{-1} and reheated at 0.1 °C s^{-1}: (a) 0.06, (b) 0.33, (c) 0.5 wt % Mg.

3.2. Microstructure

Fig. 8a–c are the microstructures obtained from samples corresponding to the three alloys (i.e. with 0.06, 0.33 and 0.5 wt % Mg, respectively), and solidified at 10 °C s^{-1}. No Mg_2Si precipitates are observed, even at the highest magnesium content. However, fine particles believed to be $Al_5Mg_8Si_6Cu_2$ are seen between the Al_2Cu crystals (arrowed). The amount of this phase is found to increase progressively with increase in magnesium content. For comparison, the microstructure of 380 + 0.5Mg alloy solidified at 0.4 °C s^{-1} is shown in Fig. 8d, where Mg_2Si precipitates are clearly seen (arrowed).

From Table IV, the start-of-melting temperature for 380 + 0.5Mg alloy heated at 0.1 °C s^{-1} lies around 492 °C. Thus, solution treating this alloy at low temperatures (i.e. 480 °C) is not expected to cause significant changes in the microstructure, Fig. 9a. Solution treatment, however, at 515 °C for 8 h results in partial melting of the Al_2Cu phase without noticeable changes in the eutectic silicon morphology, Fig. 9b. In contrast, solution treating the water-chilled samples of 380 and 380 + 0.5Mg at 515 °C for the same period of time is associated with marked changes in the silicon particle morphology, as shown in Fig. 10a and b, respectively. The black spots appearing in Fig. 10b are attributed to the fusion of the Al_2Cu phase resulting from a lowering of the start-of-melting temperature due to the presence of a high magnesium level ($\geqslant$ 0.3 wt %) in this alloy.

TABLE V Remelting characteristics of the three alloys used, solidified at 10 °C s^{-1}

Mg (wt %)	Heating rate (°C s^{-1})	I_s (°C)	(Al_2Cu[a], Mg_2Si) (°C)	Si_{eut} (°C)	β-iron (°C)	I_f (°C)	Melting range (°C)
0.06	0.1	520.0	525.90	566.49	576.5	591.8	71.8
0.33		503.5	516.78	565.45	576.5	585.9	82.4
0.50		505.9	514.51	564.22	574.1	585.9	80.0
0.06	1.0	516.5	523.88, 537.6[b]	578.34	N/A	621.2	104.7
0.33		503.5	513.17, 530.0[b]	576.39	N/A	625.9	122.4
0.50		505.9	512.50, 531.7[b]	576.27	N/A	623.5	117.6

[a] With or without $Al_5Mg_8Si_6Cu_2$.
[b] A hump instead of a well-defined peak, extending over a temperature range.

TABLE VI Comparison between the hardness values reproted by Dunn and Dickert [4] and those obtained in the present work

Mg (wt%)	Condition	Hardness (BHN)	
		[4]	Present work
0.06	As-cast	81.9	84.0
	4 h/180 °C (T5)	84.8	85.0
0.54	As-cast	89.7	101.9
	4 h/180 °C (T5)	96.7	126.0

3.3. Ageing behaviour

The role of magnesium content on the strength of 319 alloy (6.25% Si, 3.72% Cu, 0.709% Fe, 0.33% Mg, 0.3% Mn, sand mould cast, cooling rate < 1 °C s^{-1}) in the as-cast and heat-treated condition (T5 – artificial ageing at 205 °C for 8 h and quenching) has been studied by Das Gupta *et al.* [9]. Their results show that increasing magnesium levels up to 0.6% has a negligible effect on the hardness and tensile strength of both the as-cast and heat-treated alloy. Furthermore, they observed no significant microstructural changes with increasing magnesium content.

Dunn and Dickert [4] compared the effect of magnesium up to 0.55% on the mechanical properties and hardness of A380 (3.25% Cu, 1.01% Fe, 9.15% Si, 0.15% Mg, 0.43% Mn and 1.81% Zn) and 383 (2.69% Cu, 1.05% Fe, 10.5% Si, 0.035% Mg, 0.35% Mn and 2.1% Zn) alloys. The alloys were examined at three magnesium levels, i.e. 0.1, 0.35 and 0.55 wt%. The presence of magnesium was seen to increase the tensile strength, yield strength and hardness at all temperatures. Elongation was observed to be reduced by the presence of magnesium; however, the minimum value appeared to be acceptable provided the magnesium content did not exceed 0.35%. According to them, the optimum T5 treatment is 4 h at 180 °C.

The effect of varying magnesium content on the mechanical properties of a 380 aluminium die casting alloy as a function of ageing time at room temperature as well as at 180 °C was studied by Jonsson [10]. He concluded that the tensile and yield strengths increased substantially with increasing magnesium

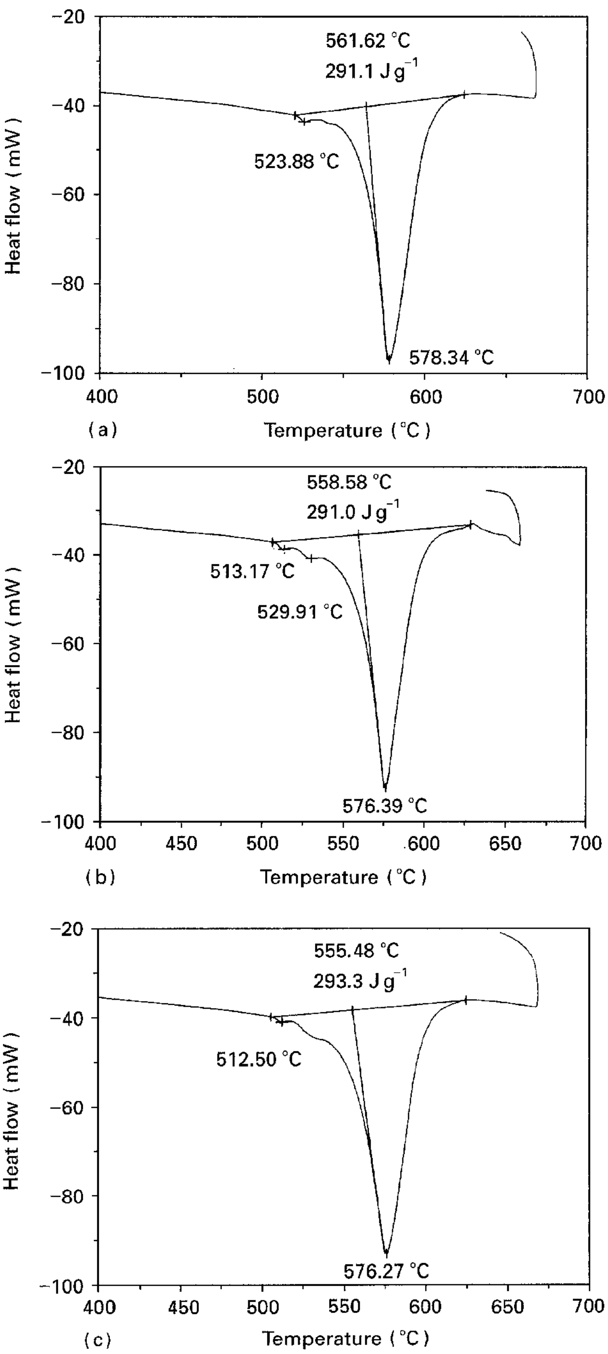


Figure 7 DSC curves of samples solidified at 10 °C s^{-1} and reheated at 1 °C s^{-1}: (a) 0.06, (b) 0.33, (c) 0.5 wt % Mg.

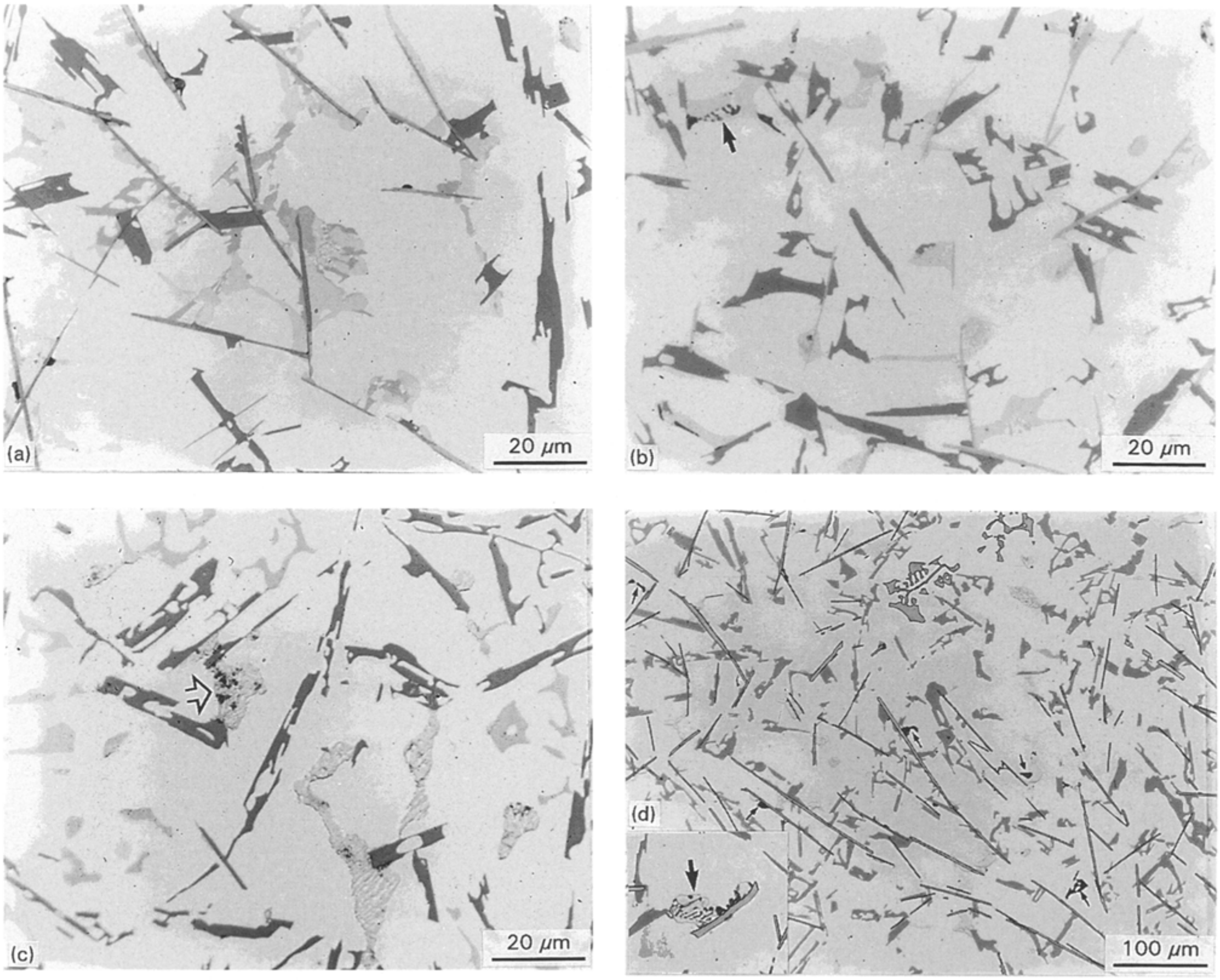


Figure 8 Optical micrographs of the water-chilled alloys: (a) 0.06, (b) 0.33, (c) 0.5 wt % Mg, (d) optical micrograph of 380 + 0.5 Mg alloy solidified at 0.4 °C s^{-1}. Inset shows the presence of $Al_5Mg_8Si_6Cu_2$ together with Al_2Cu eutectic.

content when the alloys were subjected to artificial ageing. The hardness (measured by Brinell) increased from 76 BHN to 96 BHN when 0.57 wt % Mg was added to the 380 alloy. Ageing for 8 h at 180 °C increased this value to 112 BHN, a value higher than that obtained on ageing the casting at room temperature for 1 year, i.e. 107 BHN.

The work of Klein [2] reveals that 380 die castings do not reach their peak strength at an ageing temperature of 140 °C, even after 60 h. A high rate of increase in the values of yield (YS) and tensile (UTS) strength is noted up to an ageing time of about 8 h at 160 °C ageing temperature. The values fall when the ageing time exceeds 24 h. Maximum Brinell hardness, YS and UTS are obtained for an ageing time of 18 h.

Singh *et al.* [11] proposed an empirical equation to determine the effect of (a) copper percentage, (b) magnesium percentage, (c) silicon percentage, (d) solutionizing temperature, (e) ageing temperature, and (f) ageing time on the hardness of Al–Si–Cu–Mg alloys. Their results indicate that both copper and magnesium enhance the strength properties of graphite-cast cast and heat-treated alloys. Among various interactions, those between magnesium and silicon, and magnesium, copper and silicon are most significant. However, the obtained equations are non-linear in nature, suggesting adoption of a higher order design for the prediction of hardness. The authors recommend an optimum treatment combination: solution treatment at 530 °C for 6 h followed by ageing at 185 °C for 4 h, for maximum strength properties of graphite cast alloys.

Fig. 11a–c are the plots of hardness for the 380 alloys with the three levels of magnesium, aged immediately after solidification (T5 temper) at 25, 150, 180, 200 and 220 °C for times up to 200 h. The main observations that can be made are as follows.

1. Maximum attainable hardening contribution due to the addition of 0.33 wt % Mg can be as high as 52%. Further increase in the magnesium content, i.e. up to 0.5 wt%, results in a marginal increase in the alloy strength (~ 6.5%).

2. Peak-hardening is obtained when the alloy is aged at 155 °C for 6–8 h or at 180 °C for 4 h. Prolonged ageing at 180 °C causes material softening due to overageing. A similar behaviour is observed when the alloy is aged at a higher temperature, i.e. 200 °C.

3. The hardening effect produced at 200 °C is comparable to that exhibited by the alloy upon ageing at 25 °C, indicating the commencement of overageing.

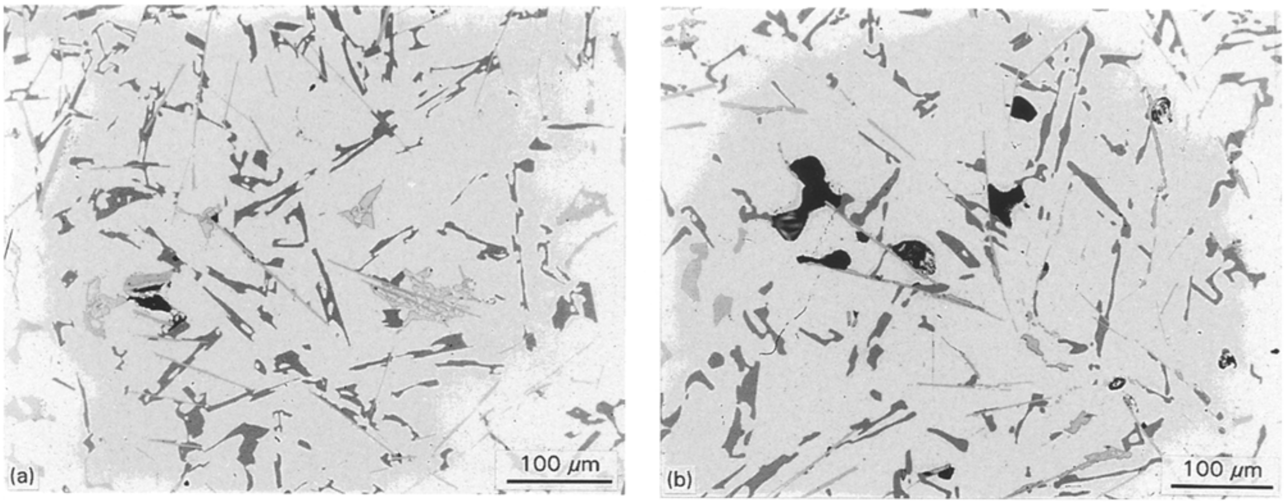


Figure 9 Optical micrographs of 380 + 0.5Mg alloy (cooling rate 0.4 °C s^{-1}), solution treated for 8 h at (a) 480 °C, and (b) 515 °C.

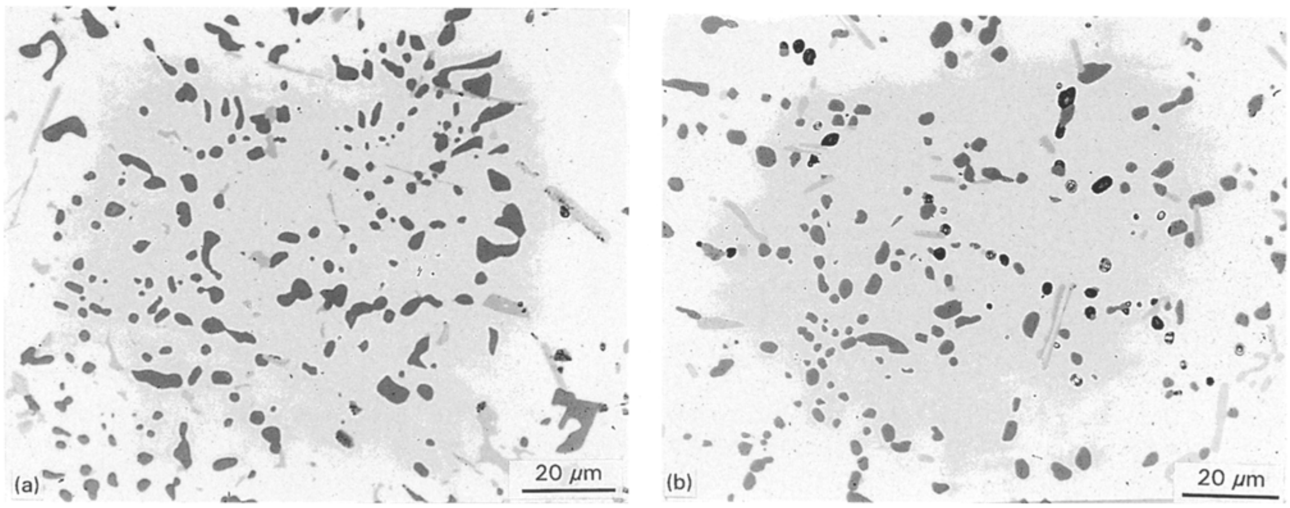


Figure 10 Optical micrographs of water-chilled samples solution treated for 8 h at 515 °C: (a) 0.06, (b) 0.5 wt % Mg.

4. Ageing at 220 °C results in material strength much inferior to that recorded immediately after casting.

The maximum hardening effect reported by Dunn and Dickert [4] when 0.57 wt % Mg was added to 380 alloy in the T5 condition (4 h at 180 °C) was less than 10% (see Table VI). The cooling rate used by them, however, was not given. Hence, it is difficult to assess the cause for the difference between their results and those obtained in the present case.

Fig. 12 displays the changes in the alloy hardness after solution treatment at 480 °C/quenching in warm water (60 °C)/immediate ageing at 180 °C for times up to 100 h (designated TT6). It is evident that there is a remarkable improvement in the alloy strength compared to that shown in Fig. 11c. When the alloy was aged 4 h in both cases, the as-cast hardness increased up to 40% compared to 16% for the T5 treated 380 + 0.33Mg alloy. Even for the base alloy, ageing for 8 h raised the as-cast hardness by 30%, whereas the T5 treatment showed an increase of almost 10%.

The work of Apelian *et al.* [12] shows that in order to obtain a maximum concentration of magnesium and silicon particles in solid solution, the solution temperature should be as close as possible to the eutectic temperature. Some of the alloy constituents may form complex eutectics, which melt at temperatures much below the equilibrium eutectic temperatures, i.e. Al_2Cu phase. The diagram of equilibrium solubility of magnesium in solid aluminium indicates that about 0.4 wt % Mg can be placed in solution at 480 °C and increases to 0.53 wt % at 515 °C. Thus, the applied cooling in the present work is not high enough to retain all magnesium in solid solution (Fig. 8). It should be noted, however, that the cooling rate used in a typical die-casting process is at least an order of magnitude greater than that employed here. Therefore, it is expected that under industrial conditions, the T5 temper could well be used to replace the T6 temper (present work), to achieve the same alloy strength.

In the second case, the as-cast samples were solution heat treated for 8 h at 515 °C/water quenched

(60 °C)/immediately aged at 155 and 180 °C for times up to 100 h (designated T6). Fig. 13a demonstrates the variation in the alloy hardness during ageing at 155 °C. The peak hardening for the base alloy is attained after 20 h, whereas that for 380 + 0.33 Mg or 380 + 0.5 Mg alloys is attained in the much shorter time of ~ 8 h. As expected, increasing the ageing temperature to 180 °C shifts the peak hardening to lower times, i.e, 8 and 4 h, respectively. Comparing Fig. 13b with Fig. 12 emphasizes the influential role of placing magnesium in solid solution on the attainable hardness level of the alloy upon subsequent ageing.

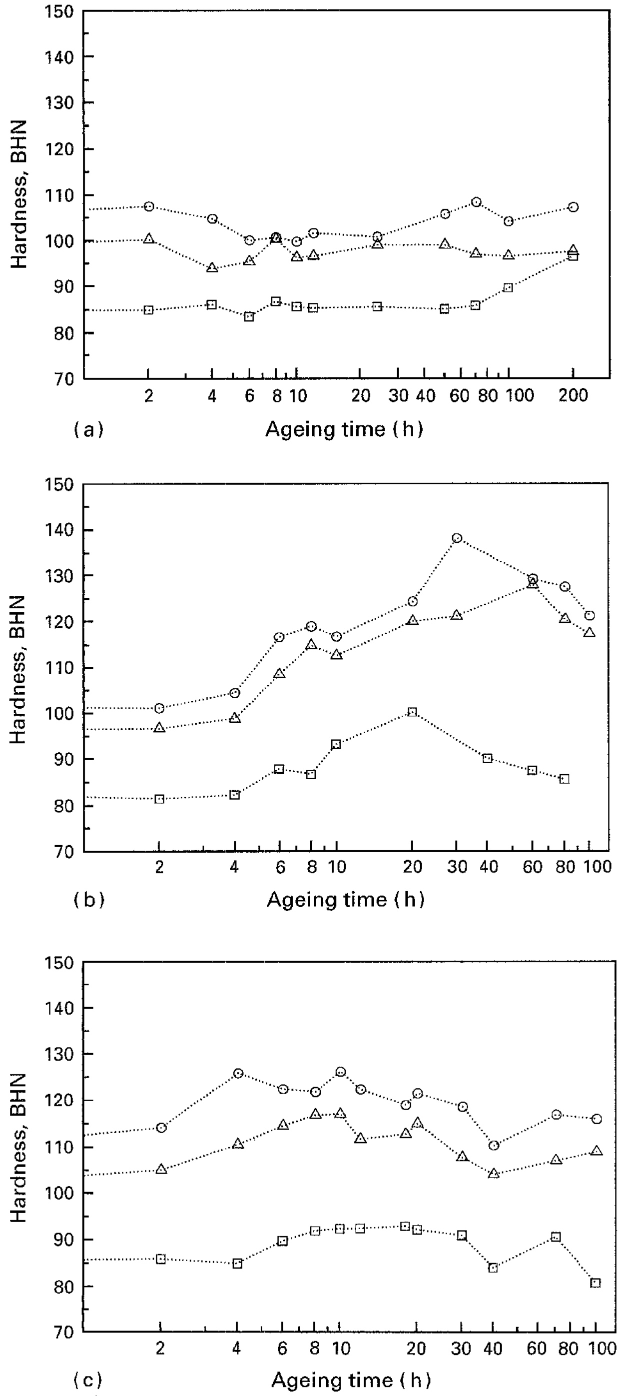


Figure 11 Variation in alloy hardness as a function of ageing time at (a) 25, (b) 155, (c) 180, (d) 200, and (e) 220 °C for (□) 0.06 wt % Mg, (△) 0.33 wt % Mg and (○) 0.50 wt % Mg.

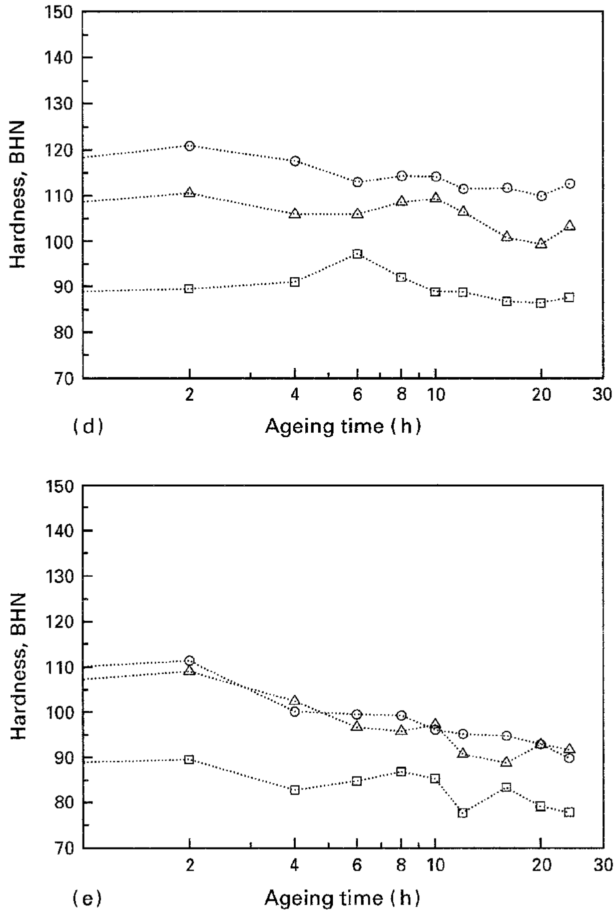


Figure 11 (continued).

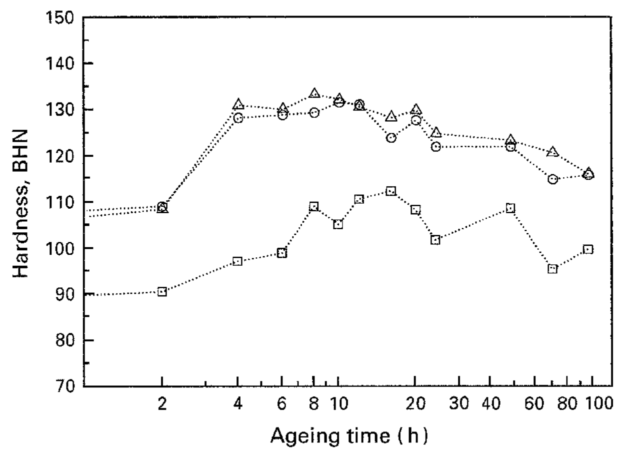


Figure 12 Variation in alloy hardness as a function of ageing time at 180 °C for samples solution heat treated at 480 °C for 8 h, for (□) 0.06 wt % Mg, (△) 0.33 wt % Mg and (○) 0.50 wt % Mg.

4. Conclusions

The effect of magnesium concentration on the ageing behaviour as measured by the hardness of water-chilled castings of 380 alloy containing 0.06, 0.33 and 0.5 wt % Mg, respectively, was studied. From an analysis of the microstructural, DSC and hardness results obtained, the following conclusions may be drawn.

1. Of the two heat treatments applied to the as-cast alloys, the significantly higher hardness values obtained with the T6 treatment can be explained by the excess precipitation of magnesium-containing phases

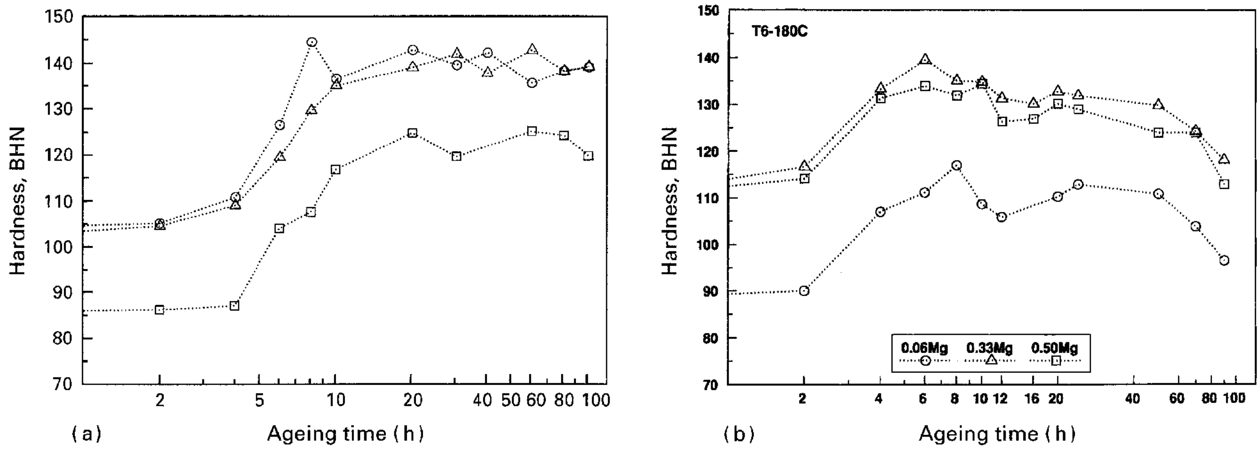


Figure 13 Variation in hardness of alloys solution heat treated at 515 °C for 8 h and aged up to 100 h at (a) 155 and (b) 180 °C, for (□) 0.06 wt % Mg, (△) 0.33 wt % Mg and (○) 0.50 wt % Mg.

in the as-solidified alloys. This excess precipitation can well be eliminated under the cooling rate conditions typical of die-casting operations, so that a T5 treatment can be employed instead of the T6 treatment to achieve the same level of improvement in the alloy strength.

2. Solution heat treatment in the low-temperature range of 480–515 °C is adequate to produce the required changes in silicon morphology and dissolution of magnesium in the aluminium matrix.

3. No significant difference in hardness behaviour is observed when the magnesium content is increased beyond 0.3 wt %.

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