The role of alloy composition in the heat treatment of aluminium high pressure die castings

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ABSTRACT

High pressure die-cast (HPDC) aluminium components that respond to age hardening cannot normally be solution treated at high temperatures because the presence of internal porosity and entrapped gases leads to the formation of surface blisters. Parts may also become dimensionally unstable due to swelling. These factors that prevent heat treatment present significant limitations to the utilisation of HPDC components. Now it has been found that blistering and dimensional change can be avoided by using modified shorter solution treatment procedures which still allow strong responses to age hardening to be achieved with a wide range of Al-Si-(Cu/Mg) alloys. In the present paper, the roles of critical alloying elements are considered in both current commercial and experimental alloy compositions in this series. It is shown that values of 0.2% proof stress exceeding 400 MPa may be readily achieved by heat treating conventionally produced HPDC components.

RIASSUNTO

I componenti in lega di alluminio ottenuti per pressocolata da rafforzare tramite invecchiamento non possono subire normalmente un trattamento di solubilizzazione ad alta temperatura in quanto la presenza di porosità interne o di gas intrappolati conduce alla formazione di rigonfiamenti superficiali. Alcune parti possono inoltre diventare dimensionalmente instabili a causa di dilatazioni. Questi fattori che contrastano il trattamento termico comportano una limitazione importante nell'utilizzo di componenti pressocolati. Si è ora riscontrato che rigonfiamenti e variazioni dimensionali possono essere evitati riducendo la durata del trattamento di solubilizzazione e consentendo così a numerose leghe Al-Si(Cu/Mg) di essere notevolmente rafforzate per invecchiamento. Nel presente lavoro è stato preso in considerazione il ruolo degli elementi leganti presenti sia nelle usuali leghe commerciali sia in leghe sperimentali. È stato evidenziato che trattamenti termici di componenti ottenuti con pressocolate tradizionali possono dare facilmente luogo a carichi di snervamento (0,2%) superiori a 400 MPa.

KEYWORDS

Heat Treatment, Precipitation, Aluminium, HPDC

INTRODUCTION

As shown in Fig. 1, high pressure die castings (HPDC's) normally retain a high gas content which arises from the presence of entrapped air, hydrogen and vaporized die lubricant. Because of this, HPDC components cannot be given a conventional solution treatment (e.g. at 520°C for 8h) as the pores containing gaseous elements may expand resulting in the unacceptable surface blisters, distortion and lower mechanical properties. The dimensions of the die cast parts may also change due to swelling with a consequent adverse effect on mechanical properties.

Recent work within the CSIRO Light Metals Flagship [1-4] has revealed a heat treatment cycle for HPDC aluminium alloys that avoids these problems. The modified

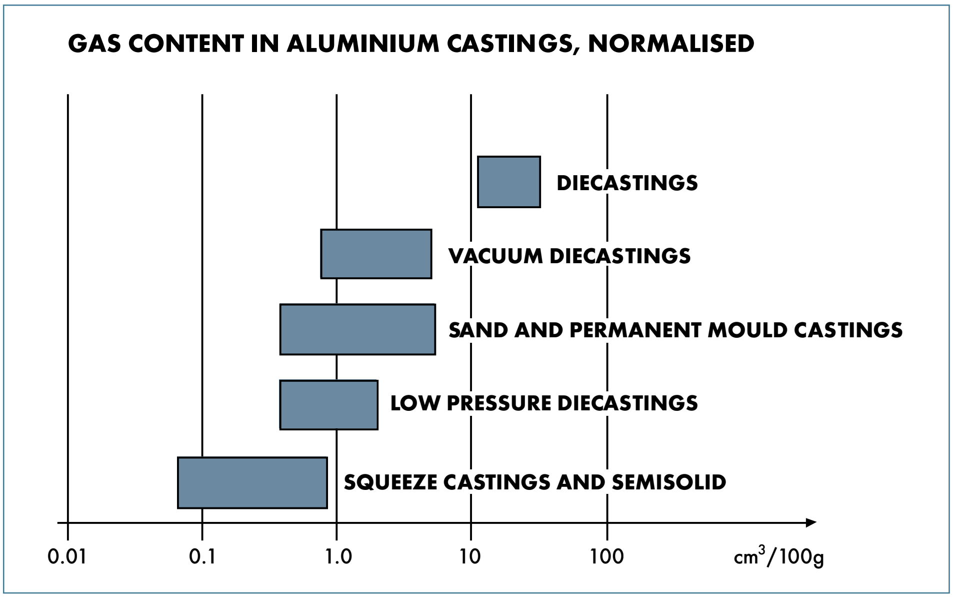

Fig. 1: Comparative gas contents arising from different casting processes [5].

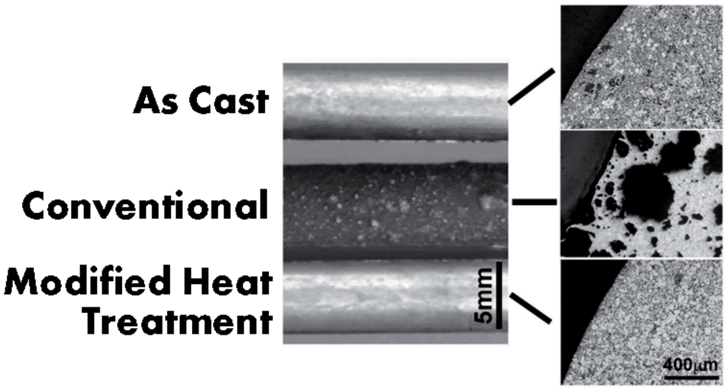

Fig. 2: Comparisons of surfaces and internal microstructures for a HPDC alloy in the as-cast, conventionally heat treated and heat treated according to the modified heat treatment conditions.

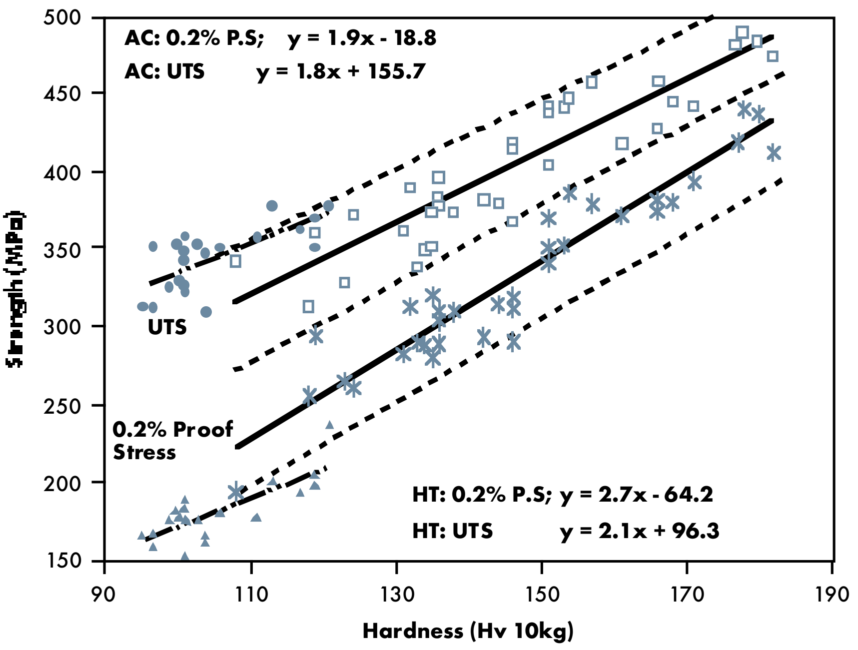

Fig. 3: Relationships between hardness and tensile properties in as-cast (solid symbols) or heat treated (open symbols) HPDC's. Each data point represents 5 or more tensile tests. All heat treated samples were artificially aged to peak strength and hardness at different temperatures. Ranges for 0.2% proof stress and tensile strength are shown as dotted lines. Equations approximating the line of best fit are also provided.

cycle involves a severely truncated solution treatment stage at lower than normal temperatures to avoid blistering, which, due in part to the unique microstructure generated by the high pressure die-casting process, is sufficient to attain at least a partial solid solution of the potent age hardening elements Cu, Mg and Si. Comparisons between the surface finish and internal microstructures of high pressure die-cast test components in the as-cast, conventionally solution treated, or solution treated according to the new procedures are shown in Fig. 2. Following solution treatment and quenching, the HPDC alloys may be naturally aged to a T4 temper at ambient temperature, or artificially aged at an intermediate temperature (e.g. 150-180°C). As a result, large increases in tensile properties may be realized and a summary of hardness and strength values of heat treated HPDC's compared with those for the as-cast alloys is shown in Fig. 3. A T4 temper normally raises the 0.2% proof stress by an average of 30% above the as-cast condition, often with elevated levels of ductility. Ageing to a T6 temper allows peak strength to be achieved and values of 0.2% proof stress may be raised by 80 to 100% or more.

With respect to their potential for strengthening by heat treatment, normal Al-Si-(Cu/Mg) HPDC alloys contain amounts of the critical elements Si, Mg and Cu that contribute to age hardening. Although up to 3% Zn is permissible in some HPDC alloy variants, there is often insufficient Mg present to form substantial quantities of MgZn2 precipitates that could also contribute to strengthening. The current paper is concerned with the roles of the various alloying elements in contributing to age hardening in a wide range of HPDC alloys.

EXPERIMENTAL METHODS

HPDC alloy specimens for tensile testing were produced using a Toshiba horizontal cold chamber die-casting machine with a 250 tonne locking force, a shot sleeve with an internal diameter of 50mm and a stroke of 280mm. The die provided two cylindrical tensile specimens and one flat tensile specimen from each shot, and each conformed to specification AS1391. The cylindrical tensile test bars used for the current work had a total length of 100 mm, with a central parallel gauge section 33 mm long and a diameter of 5.55±0.1mm. The runner and samples produced are shown in Fig. 4. Five tensile samples were tested in each condition and ten alloy compositions were examined which are listed in Table 1. Levels of Cu, Mg, and Zn were varied within the specified composition ranges to assess their effects on mechanical properties of the heat treated alloys.

Tensile properties were determined for all of the alloys in each of the three conditions examined (as-cast, T4 and T6 tempers). Solution treatment for all heat treatments was standardized at 490°C for 15 minutes

Fig. 4: The runner and test specimens produced for the current evaluation.

total immersion time in a circulating air furnace, followed by water quenching. This procedure has been shown elsewhere to avoid blistering and still produce a high response to age hardening in these test specimens [1-4]. Samples for the T6 tempers were subsequently aged in oil at 150°C for 24h whereas those for T4 tempers were aged at 25°C in air for a standard time of 14 days.

Table 1. Compositions of ten alloys examined for heat treatment

Alloy	Al	Si	Cu	Mg	Zn	Fe	Mn	other
1	balance	9	3.1	0.1	0.53	0.86	0.16	<0.2
2	balance	9.1	3.18	0.29	0.6	0.86	0.14	<0.2
3	balance	8.6	3.6	0.1	0.53	0.93	0.18	<0.2
4	balance	8.6	3.6	0.3	0.53	1.0	0.2	<0.2
5	balance	8.5	4.9	0.1	0.51	0.97	0.2	<0.2
6	balance	8.7	4.9	0.27	0.51	1.0	0.21	<0.2
7	balance	8.8	4.0	0.7	0.56	1.1	0.19	<0.2
8	balance	9.1	2.04	0.26	0.53	0.79	0.36	<0.2
9	balance	9.2	3.11	0.09	2.9	0.9	0.16	<0.2
10	balance	9.1	4.2	0.22	1.2	1.3	0.2	<0.25

Transmission electron microscopy was conducted on selected alloys to observe precipitation occurring within the aluminium grains. TEM discs were prepared by standard polishing techniques, and examined in a Jeol 2000EX microscope operating at 200 kV.

Table 2. Mechanical properties of 10 alloys examined tested in the as-cast, T4 or T6 temper

Alloy	1	2	3	4	5	6	7	8	9	10
As Cast	172,	189,	176,	200,	193,	206,	237,	157,	165,	198,
	354,	358,	358,	362,	363,	369,	377,	311,	347,	351,
	4.1%	3.2%	3.7%	3.0%	3.3%	2.9%	2.2%	3.8%	3.8%	2.5%
T4	217,	246,	234,	258,	244,	275,	261,	192,	224,	267,
	387,	411,	397,	411,	412,	418,	413,	306,	405,	405,
	5.8%	5.2%	4.8%	4.3%	5.5%	3.8%	3.9%	3.3%	7%	3.4%
T6	352,	374,	379,	419,	381,	440,	413,	291,	370,	437,
	441,	458,	457,	474,	446,	490,	474,	368,	442,	484,
	3.4%	2.4%	2.7%	2.2%	1.9%	1.6%	1.6%	2.6%	3%	1.6%

Results are shown as 0.2% proof stress (MPa), tensile strength (MPa) and elongation at failure (%).

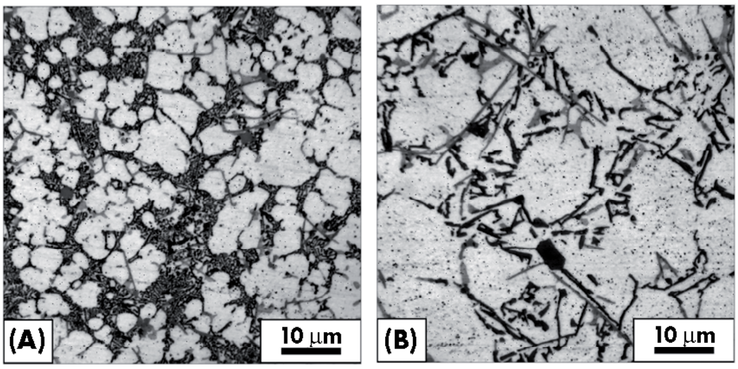

Fig. 5: Microstructures of as-cast HPDC Alloy 1 (a) at the sample edge and (b) at the sample centre. Light grey needles and blocky phases are Fe-based intermetallics.

RESULTS AND DISCUSSION

TENSILE PROPERTIES

The tensile properties of the 10 alloys examined in the as-cast, as solution treated, T4 and T6 tempers are presented in Table 2, with reference to numbered compositions listed in Table 1. As-cast values of 0.2% proof stress ranged from 157 MPa to 237 MPa. Typically, ageing to the T4 temper raised the 0.2% proof stress by about 30% above the as-cast condition, often with elevated levels of ductility. T4 0.2% proof stress values generally ranged between 190 MPa for the low Cu Alloy 8, up to 275 MPa for the high-Cu Alloy 6. Ageing to a T6 temper increased the as-cast values of 0.2% proof stress by 75 to 120%, with values that ranged from 291 MPa for the low Cu Alloy 8 up to 440 MPa for the high-Cu Alloy 6. Raising Mg levels to higher amounts (e.g. 0.7% Mg in Alloy 7) proved to be disadvantageous when compared to similar alloys with lower Mg contents (e.g. 0.22% Mg in Alloy 10).

ROLE OF ALLOYING ELEMENTS

Silicon

Silicon is present in casting alloys largely to facilitate high fluidity in the molten state, to form a eutectic, and to promote the formation of strengthening precipitates through the expected reaction of Si and Mg present in solid solution. Silicon also forms intermetallic sludge particles with Fe and Mn in the melt (e.g. ,-AlFeSi5). As the Al-Si binary system displays an abnormal eutectic, solidification involves the formation of Si particles in the solidified melt. For HPDC alloys, the Si plate size varies greatly between the edge and centre of a part, because the rapid cooling rate at the surfaces of components produces an extremely fine aluminium grain size and Si particle size, such as shown in Fig. 5(a). Towards the centre of a component, the Al grain size is larger and the Si particles are present in fewer numbers and larger sizes (Fig. 5(b)).

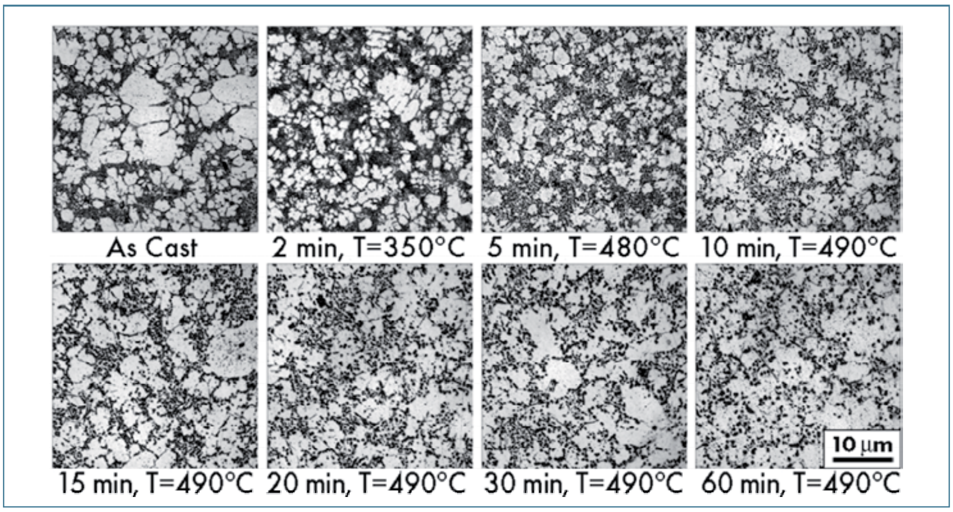

Fig. 6: The refinement of silicon particles during heating to 490°C. Note surface blisters form at times of ?20 minutes. See also Fig. 7 for quantification.

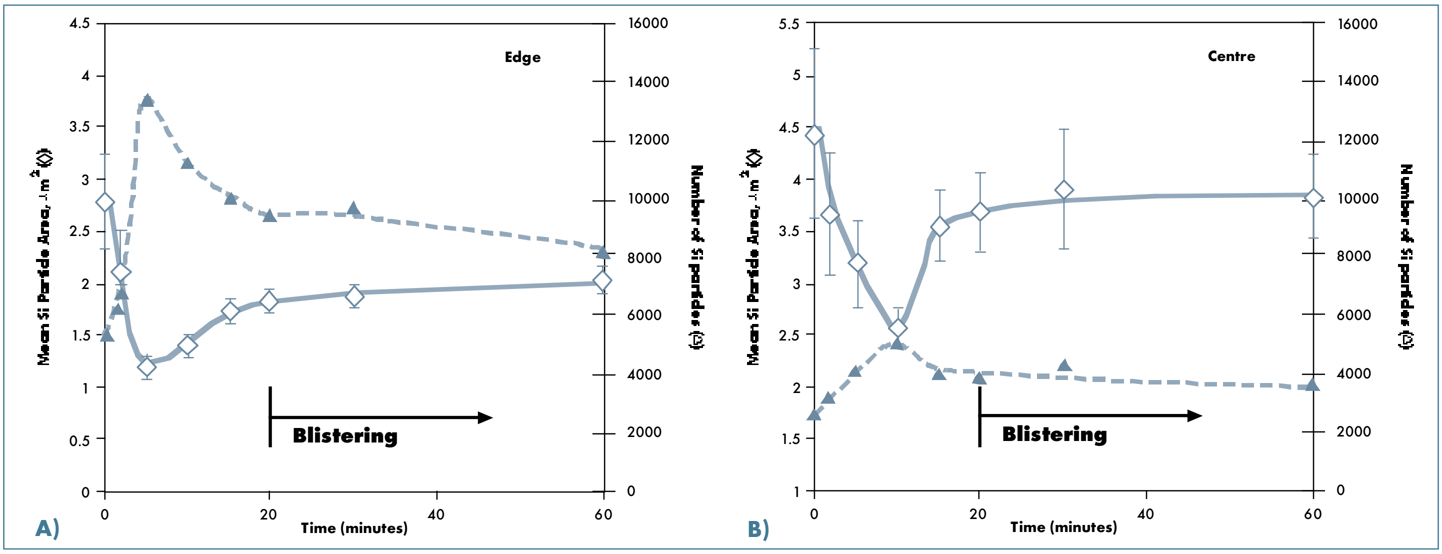

Fig. 7: Quantification of the changes to silicon particle size (diamonds) and number (asterisks), for (a) edge and (b) centre regions of the die-castings. Total area examined for each data point is ~122000Ìm2. The silicon particles fragment and spheroidize, then undergo Ostwald ripening, decreasing in number.

During heating of the HPDC alloy to 490°C, the Si present in the solidified eutectic rapidly undergoes significant morphological changes; the Si particles fragment and then quickly spheroidize. Examples of this are shown in Figs. 6 and 7. Fig. 6 shows the microstructural changes to the silicon particles at the edge of samples, and Fig. 7 quantifies these changes for both edge and centre regions. It may be noted that most of the changes to the Si particles occurred during an immersion time of less than 20 minutes, which corresponds closely to the appropriate heat treatment time used to avoid blistering on the surfaces of samples. Fig. 8 shows the microstructure of a sub-surface region that

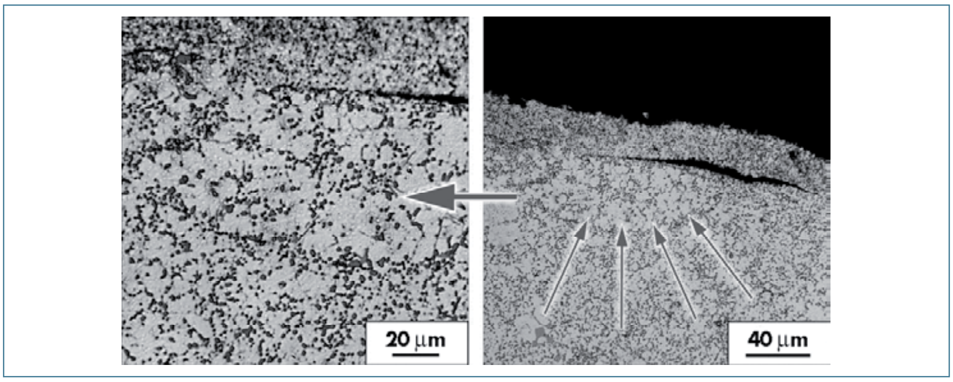

Fig. 8: A laminar defect leading to blistering in an A380 alloy, solution treated 20 minutes at 490°C. A similar, unexpanded laminar defect is also shown arrowed and at higher magnification. Note the distribution of pinning silicon particles (arrowed at right) along the defect length.

contains a blister under which there appears to be a thin laminar defect decorated with small Si particles (arrowed). It may be that these particles have so far served to pin this defect thereby preventing its expansion into an actual blister. Such features have been observed to be discontinuous defects displaying only partial metallurgical bonding across their interfaces. These appear to be responsible for a significant proportion of blisters that are generated when samples are held for longer times or at higher temperatures. A simple explanation is that small pockets of gas entrapped at high pressure in these laminar defects expand to form small cracks. These cracks can then extend and coalesce with similar cracks forming in other gas pockets so that they produce large surface blisters like the one shown at an early stage of formation in Fig. 8.

The role of Si on precipitation processes during the age hardening of many HPDC's is quite complex, as it is involved in precipitation associated with the complex Q? phase, Al5Cu2Mg8Si6 or the precursor L phase and will be discussed later.

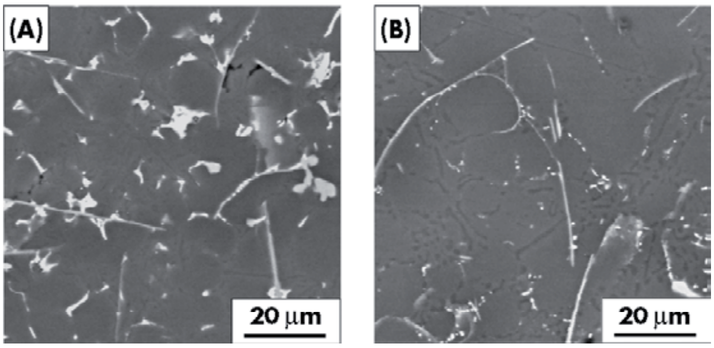

Fig. 9: (a) As cast and (b) T6 treated HPDC Alloy 1. Despite the sort duration of the solution treatment step, the Cu-containing phases are largely dissolved leaving remnant Fe and Mn-bearing phases.

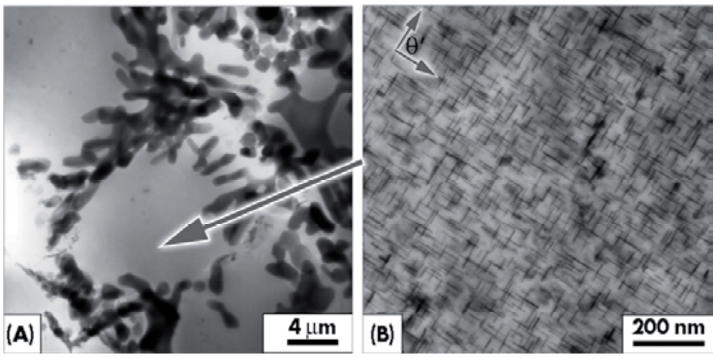

Fig. 10: TEM micrographs of HPDC A380 Alloy 1 in the T6 temper. (a) shows the α-aluminium grains bounded by the silicon phase and (b) finely dispersed θ' precipitates within the α-aluminium grains in an $[001]_{\alpha}$ orientation.

Copper

The presence of Cu in HPDC Al-Si-Cu alloys may also assist fluidity by depressing the solidus temperature of the alloy and die-casting experiments have revealed that Alloys 5 and 6 are particularly well suited to production of thin walled castings. Copper also allows as-cast HPDC's to develop reasonable strengths due to the limited precipitation that occurs during cooling from the casting temperature. During solution treatment, Cu dissolves rapidly into the aluminium matrix despite the short duration employed (Fig. 9) and this element is critical in facilitating age hardening, particularly when Mg and Si are also present.

The strengthening that Cu additions provide to the Al matrix is caused by precipitation of solute clusters and fine Guinier-Preston (GP) zones in the T4 condition, or as other phases such as θ (Al_2Cu) or Q' ($Al_5Cu_2Mg_8Si_6$) that form in the T6 condition (e.g. Fig. 10). In Fig. 10 it is evident that the fine dispersion of precipitate plates present within the aluminium grains is similar to that observed in binary Al-Cu wrought alloys. There is, however, a large difference in the strength that develops when binary wrought and HPDC alloys containing similar amounts of Cu are age hardened (Table 3). In each case, the alloys were aged to peak strength at 150°C (T6 temper) and the 0.2% proof stress of the more complex HPDC alloy is 143 MPa higher. This difference can be attributed mainly to the presence of the spheroidized particles of Si that provide substantial dispersion hardening.

The tensile properties of Alloys 2,5,7&9, which contain similar levels of Mg and varying levels of Cu, were determined for the as-cast, T4 and T6 tempers and the results are summarized in Fig. 11. Here it may be seen that increasing Cu content improves the strength properties of the die-castings in all three conditions. For the as-cast alloys, the 0.2% proof stress increased linearly between 2 wt% and 3.6 wt%, with no further improvements being evident when the Cu was raised further to 4.9 wt%. The tensile strength of these alloys increased proportionately to the 0.2% proof stress. Ductility was little changed with only a slight decrease occurring as the Cu content was raised from 2 to 4.9%. For the T4 temper, the 0.2% proof stress increased significantly as the Cu content was raised from 2 to 3.2wt% and then at a slower rate for higher levels of Cu. In all alloys except Alloy 8, ductility in the T4 temper was in general higher than those for the as-cast condition. Another observation is that, for Alloy 6 (Cu 4.9%) the 0.2% proof stress for the T4 temper was 275 MPa which is similar to that for the Al-Si-Mg (permanent mold or sand cast) alloy A357 aged to a T6 temper [e.g. 6].

Alloys aged to a T6 temper showed the highest strength levels. The 0.2% proof stress for Alloy 8 (2%Cu) was 291 MPa and rose linearly with Cu content to 419 MPa for Alloy 4 (3.6%). Further small increases occurred as Cu level was raised to 4.9%.

Table 3. A comparison of the properties developed from either binary Al-Cu alloy or the more complex HPDC alloy, displaying similar Al:Cu ratio

Product form	Alloy (wt%)	Alloy (at%)	Atomic Ratio Al:Cu	0.2% Proof stress (MPa)	Tensile strength (MPa)	Elongation (%)
Wrought T6	Al-4Cu	Al-1.74Cu	56.5:1	236	325	5%
HPDC T6 (Alloy 3)	Al-8.6Si-3.6Cu-0.93Fe-0.1Mg-0.53Zn-0.18Mn	Al-8.54Si-1.58Cu-0.46Fe-0.11Mg-0.23Zn-0.09Mn	56.3:1	379	457	2.7%

Levels of tensile strength followed the same general trends. As with the as-cast condition, ductility for alloys aged to a T6 temper decreased gradually as the Cu content was raised.

In summary, the differences in levels of 0.2% proof stress between alloys containing 2% and 3.6% Cu was 43 MPa for the as-cast condition (increase of 27%), 66 MPa for the T4 temper (34% increase)

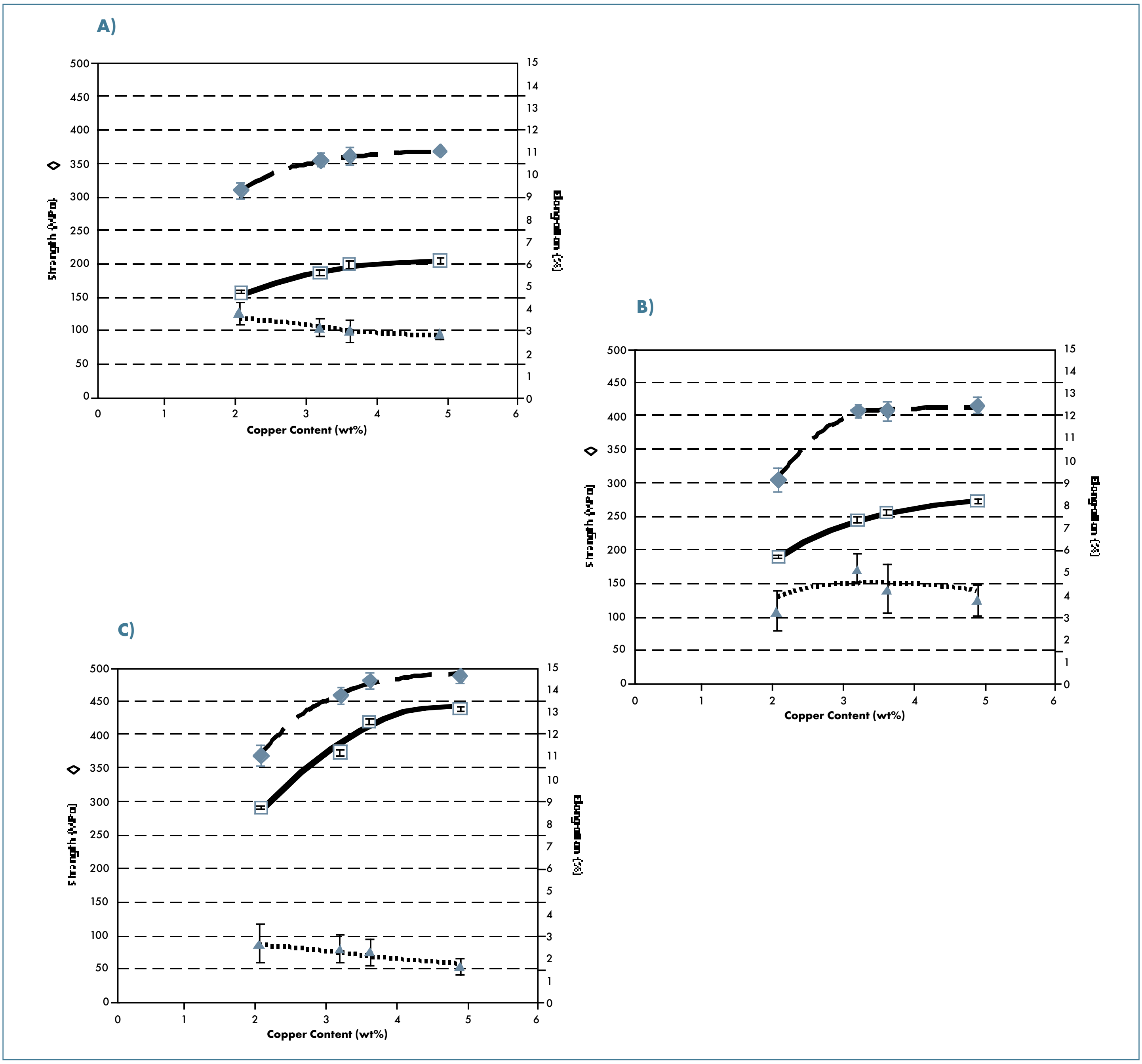

Fig. 11: Effect of increasing Cu content on mechanical properties of Al-Si-Cu HPDC's containing ~0.3Mg content and tested in (a) as-cast, (b) T4 or (c) T6 conditions. Squares show 0.2% proof stress, diamonds tensile strength and triangles elongation at failure. One standard deviation is shown on each data point. See text for details.

and 128 MPa for the T6 temper (44% increase). Ageing most alloys to a T6 temper more than doubled their as-cast levels of 0.2% proof stress.

Magnesium

When present in combination with Cu and/or Si, Mg is also a very efficient alloying addition for strengthening aluminium alloys, although in 380 type HPDC alloys it is often restricted to less than 0.3%, and even below 0.1% for some variants. In the related alloy AlSi9Cu3(Fe) (European designation), up to 0.55% Mg may be present whereas in alloy SC84R (Canadian designation), up to 0.75% is permissible. Even small amounts of Mg can have a profound effect on age hardening. The hardness-time curves for Alloys 1 and 2 aged at 150°C following solution treatment at 490°C for 15 minutes are shown in Fig. 12. Each has a similar shape but Alloy 2, which has the higher Mg content of 0.3%, shows rapid early hardening at ageing times up to 1hr. In general, the greatest benefit to tensile properties arises when the levels of Cu and Mg are raised together (see Tables 1&2) although increases in Mg above 0.3% appear to have little effect on tensile properties (at least when Cu level is high). In fact, comparison of results for Alloys 7 & 10 suggests that higher Mg contents may be detrimental to the development of improved tensile properties because increasing the Mg content from 0.22 to 0.7% has led to a decrease of 5.5% in 0.2% proof stress.

The role of Mg on precipitation processes is discussed further below, in relation to the combined effects of Si, Cu and Mg.

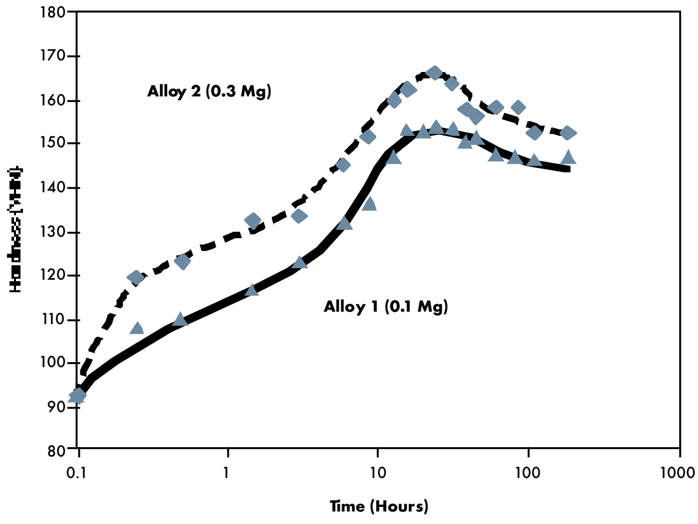

Fig. 12: The effect of Mg on the age hardening response of Al-Si-Cu-(Fe) alloys. Alloys were solution treated for 15 minutes at 490°C, quenched, then aged at 150°C. The major difference between the ageing curves occurs within the first hour, where rapid early hardening occurs for the alloy containing 0.3% Mg.

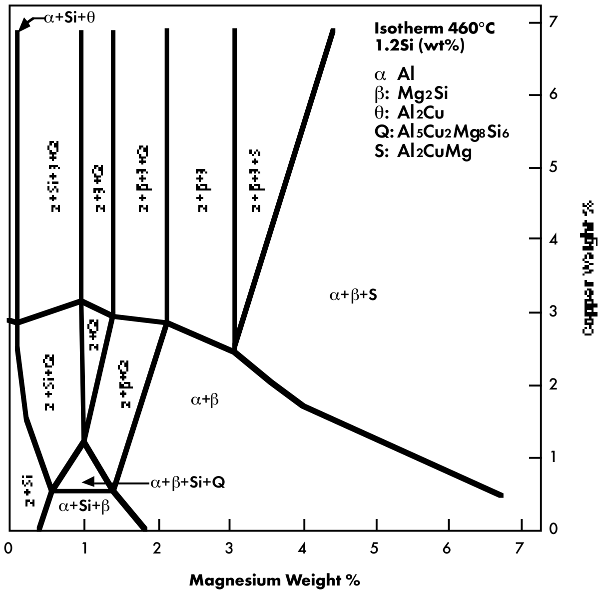

Fig. 13: The quaternary Al-Si-Cu-Mg phase diagram (isothermal section at 460°C) for a constant 1.2wt%Si [7].

Combined Roles of Silicon, Copper and Magnesium

In relation to the precise roles of Cu, Mg and Si during strengthening of HPDC's by age hardening, it is also important to consider the solute contents of the aluminium grains separately from the bulk alloy composition. Fig. 13 shows the quaternary Al-Si-Cu-Mg phase diagram at 460°C for a Si content of 1.2% [7]. If it is assumed that all Cu, ~1.2Si, and all Mg, are present in the solid solution then it is possible to predict the phases that may form during heat treatment of HPDC's. The vast majority of HPDC alloy compositions fall within regions of the diagram in which it can be predicted that the complex Q phase, or its precursors Q′ or L will form instead of θ′, β″ or S which are better known as strengthening precipitate phases. For example, Fig. 14 shows examples of Alloys 3, 4 and 8, heat treated to a peak aged condition at 150°C. Each of the alloys displays a different precipitate structure despite the only minor differences in composition. For the alloy containing only 0.1%Mg, (Alloy 3) the majority of the precipitates present within the microstructure can be identified as θ′. As predicted by the phase diagram, for the similar alloy containing 0.3Mg (Alloy 4), most of the precipitates present within the microstructure are still the θ′ phase, but now a substantial amount of fine precipitates of the Q′ phase, or it's pre-

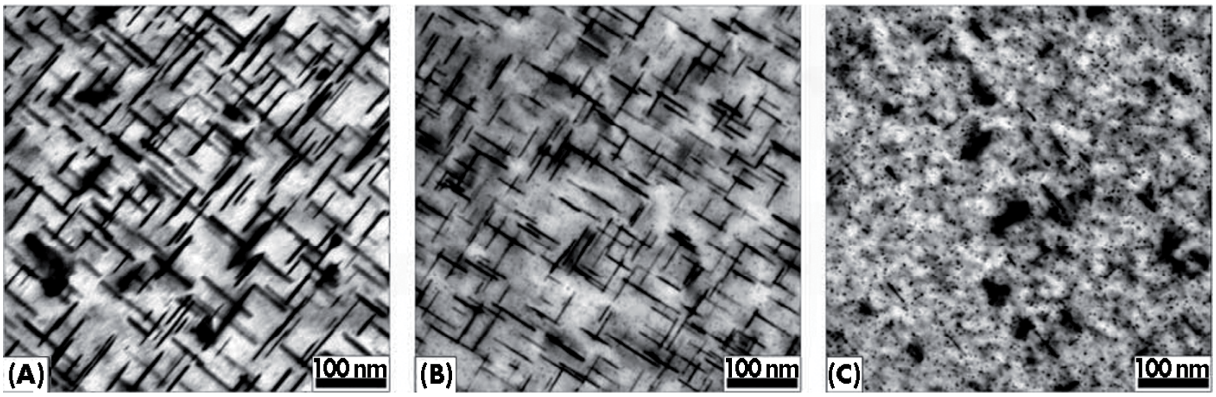

Fig. 14: Peak aged T6 precipitation within the aluminium grains. (a) is of the low- Mg Alloy 3, (b) is of the higher-Mg Alloy 4, and (c) is of the higher-Mg, lower-Cu Alloy 8. TEM images $[001]_{\alpha}$.

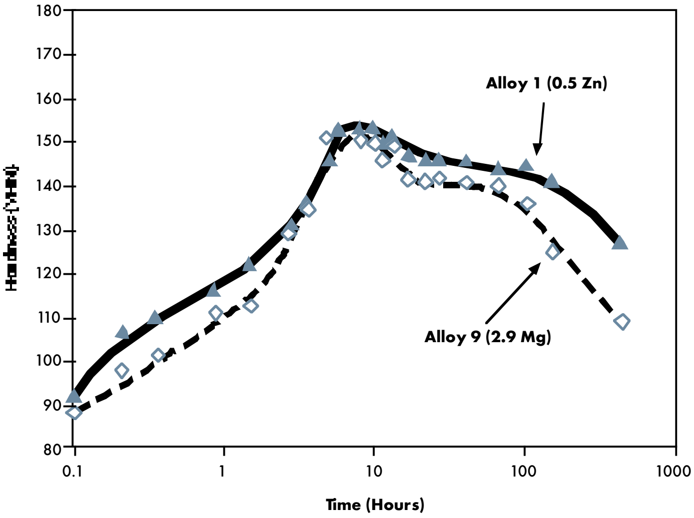

Fig. 15: Comparison of hardness-time curves for Alloy 1 (standard) and Alloy 9 (high zinc) showing the effect of raised Zn content on overageing behavior. The alloy containing 2.9 wt% Zn overages more during prolonged exposure at 150°C.

cursor L, can be detected in the background (viewed as fine, often rectangular dots). Alloy 8, which has less Cu present, forms fewer θ′ precipitates together with the very fine, dense dispersion of the L or Q′ phases throughout the matrix. As expected from Fig. 13, the mode of decomposition of the supersaturated solid solution varies with alloy composition. For example, where Cu content is reduced further and Mg content increased, the alloys are expected to precipitate a combination of the Q′ phase and β″ phase (including its GP zone and cluster precursors). As Cu is reduced still further, the relative proportions continue to change, and the dominant precipitate phase becomes β″. The most important observation, however, is that the Q′ phase or its precursor L are the dominant precipitating phases in most HPDC compositions. This also helps to explain the primary difference in (early stage) hardening between Alloy 1 and 2, (Fig. 12). This follows because, whereas θ′ is the dominant precipitating phase in the aluminium grains of Alloy 1, Alloy 2 has present a combination of L/Q′ laths that precipitate first, together with θ′ that forms later in the ageing cycle.

Zinc

Zinc is normally added to HPDC alloys to improve castability, machinability, and corrosion resistance [8] and up to 3% is permitted in some compositions. Whereas Zn and Mg together can cause significant age hardening through the precipitation of $MgZn_2$ in many aluminium alloys (e.g. 7000 series wrought alloys), the levels of Mg in the present alloys are generally low and the effect on tensile properties is very small. The increase in tensile properties that is facilitated by increasing Zn content in low Mg alloys is therefore more likely to arise from solution strengthening rather than from precipitation. High additions (i.e. ~3%) of Zn do appear to increase softening during overageing (Fig. 15).

Tin

Tin may be present in HPDC alloys due to residual contamination arising from the use of secondary metal. In general, Sn may increase fluidity during casting and improve machinability, so that specifications for HPDC compositions worldwide include varying allowances for Sn, usually up to 0.35%. Sn however may display deleterious effects during die-casting and it is recommended that its content is kept to below 0.05%. This follows because additions as low as 0.08% have been observed to cause severe flashing, die sticking and hot tearing. These complications may result in reduced productivity, greater die wear, an increased reject rate and variable part quality in the as-cast state.

Corrosion may also be adversely affected when Sn is present [8]. Sn may also act as an Fe modifier, producing large rosette shaped Fe bearing sludge particles (Fig. 16), which act as preferred sites for crack propagation. It may be noted, however, that trace additions of Sn (e.g. 0.05%) are known to stimulate age hardening in Al-Cu alloys [e.g. 9]. It was decided therefore to make a preliminary study of the effects on heat treatment of adding a small amount (0.08%) of Sn to one HPDC alloy (similar to Alloy 1).

Tensile properties were determined after either a standard T4 treatment at 25°C, or a T6 ageing treatment at 150°C. Generally, addition of Sn in binary Al-Cu alloys reduces the T4 ageing response and increases the response to T6 ageing. However, as shown in Table 4, neither effect was observed in the HPDC alloy. Therefore, since Sn was found to provide no benefit in stimulating an increased response to age hardening, it was decided that levels should be kept as low as possible for the reasons explained above.

Transition Metal Elements; (Fe, Mn, Cr)

Iron and manganese are present in HPDC alloys to minimize die soldering so that die life is increased and productivity is improved. In general, the Fe content needs to be above 0.5% which is fortunate because >95% of HPDC alloys are made of secondary metal in which Fe contents may be relatively high. Fe, Mn and Cr tend to form intermetallics that will be present in differing amounts and morphologies, depending on the so-called "sludge factor" of the alloy [10]. A proportion of sludge particles form in the melt, some in the shot sleeve, and some during solidification in the die. Generally, a simple rule that is followed by diecasters is that:

Sludge factor (SF)=1x wt%Fe + 2x wt%Mn + 3x wt%Cr ,where SF<1.7 is generally considered to be most acceptable in practice (e.g. Table 1).

The morphology of the sludge particles is strongly influenced by the Fe:Mn:Cr ratio of the alloy. Because of this, Mn and Cr are sometimes referred to as Fe correctors since they can eliminate a proportion of undesirable, needle-like, Fe-containing intermetallic phases from forming within the alloy, instead causing them to form as more innocuous particles. Other transition metal elements (e.g. Co, Ni, V) may also serve as Fe correctors and contribute to sludge formation, but Mn is the most commonly found in commercial alloy products since it is often introduced when secondary metal (e.g. canstock) is recycled. When the ratio of Fe:(Mn+Cr) is high, the sludge particles tend to be needle-like whereas when the ratio is low, the particles tend to exhibit more Chinese script, blocky, star-like rosettes or polyhedral particles. Examples of different Fe-containing particles present in die-cast microstructures are shown in Figs. 5,9&16.

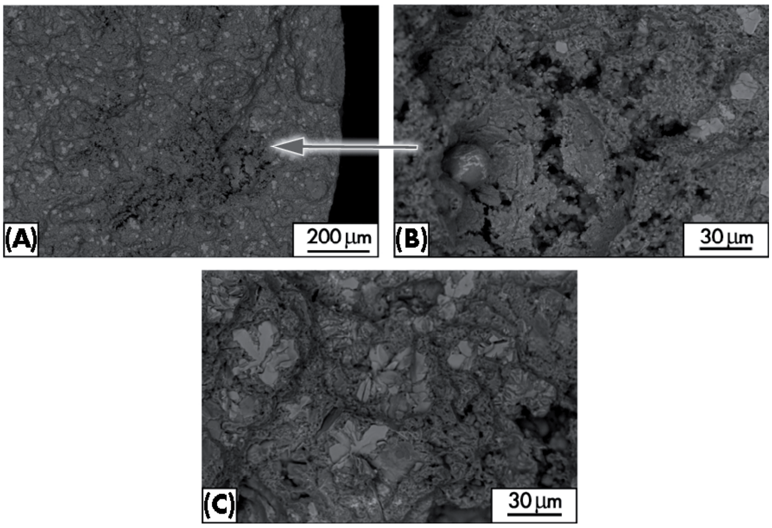

Fig. 16: Fractography of a HPDC Al-Si-Cu-(Fe) alloy containing 0.08wt%Sn. Sn even in very low amounts causes severe die sticking, hot tearing (example in (a) shown at higher magnification in (b)) as well as acting as an Fe modifier. Sn causes the formation of rosette shaped Fe-bearing particles (c), which are brittle and therefore preferred crack propagation sites. In (a), all the bright phases seen are the rosetteshaped Fe-bearing particles. Backscattered SEM.

Table 4. The effect on Sn on the ageing response of Al-Si-Cu HPDC's

	As Cast properties			T4 properties			T6 properties		
Alloy	0.2% P.S	UTS	%El.	0.2% P.S	UTS	%El.	0.2% P.S	UTS	%El.
380 alloy + 0.01Sn	176 MPa	329 MPa	2.9%	239 MPa	368 MPa	3.2%	386 MPa	448 MPa	1.7%
380 alloy + 0.08Sn	182 MPa	323 MPa	2.6%	242 MPa	363 MPa	3.0%	382 MPa	428 MPa	1.4%

The transition metal elements have no distinct role in heat treatment. Rather they cause some increase in as-cast and high temperature strength as well as improving wear and creep resistance. Because of a propensity to form brittle phases, they may have adverse effects on fracture properties and fatigue resistance.

CONCLUSIONS

1. High pressure diecast aluminium alloys that respond to age hardening may be successfully heat treated without causing blistering, by solution treating for much shorter times and at lower temperatures.
2. When heat treated to a T4 temper, HPDC's generally develop increases in 0.2% proof stress of around 30% compared to the as-cast condition, together with simultaneous improvements to ductility.
3. In a T6 temper, HPDC's generally develop increases in 0.2% proof stress

of between 75 and 120% when compared to the as-cast condition, usually with little change to ductility. The high properties developed compare very favorably with those observed for heat treated aluminium permanent mold castings and many wrought products.

4. During solution treatment, Si rapidly undergoes significant morphological changes. Initially it fragments before spheroidizing and then grows due to Ostwald ripening. When the number of Si particles is high, internal gas-containing laminar defects tend to be pinned which appears to prevent or delay their expansion that can lead to surface blistering. Additionally, the changes to the silicon morphology appear, in part, to be responsible for the improved strength/ductility relationships in the heat treated conditions.
5. Spheroidized Si particles provide significant dispersion strengthening to HPDC aluminium alloys.
6. Cu is the primary element for promoting a strong response to heat treatment in HPDC aluminium alloys. For the compositions studied that contain ~0.3% Mg, the properties continue to increase on addition of up to 4.9%Cu, whereas for lower levels of Mg (~0.1%), the properties are little changed beyond about 3.6%Cu.
7. Magnesium has a strong effect in increasing the tensile properties of heat treated HPDC Al-Si-Cu alloys. In particular, increasing the Mg content from 0.1% to above 0.2% appears to initiate rapid early hardening. However, increasing the Mg content above 0.3% and up to as much as 0.7% causes no further improvement in tensile properties. Where Cu content is 3.6% or above and Mg content is >0.2%, 0.2% proof strength values may exceed 400 MPa.
8. Zinc additions have little effect on increasing the tensile properties of aged HPDC aluminium alloys, providing only minor increases in strength via solution strengthening, because the Mg content is low. The element has the disadvantage of accelerating softening and overageing at elevated temperatures.
9. Tin is regarded as having a deleterious effect because it causes die soldering, flashing and hot tearing in HPDC aluminium alloys, as well as promoting the formation of large, Fe-bearing sludge particles in the microstructure. Since it has been found that this element has no effect on the response of the Al-Si-Cu alloys to age hardening, it is recommended that its content be kept as low as possible.
10. Transition metals present in HPDC alloys have no significant effect on the response to ageing. These elements form brittle intermetallic compounds and as a result may have an adverse effect on fracture and fatigue properties.

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