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Research Paper

"HEAT TREATMENT OF AL-SI-CU-MG CASTING ALLOYS FOR THE MANUFACTURING OF LIGHT WEIGHT MACHINES/ VEHICLE PARTS WITH INCREASED STRENGTH."

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Environmental savings can be made by increasing the use of aluminum alloys in the automotive industry as the vehicles can be made lighter. Increasing the knowledge about the heat treatment process is one task in the direction towards this goal. The aim of this work is to investigate and model the heat treatment process for Al-Si casting alloys. Three alloys containing Mg and/or Cu were cast using the gradient solidification technique to achieve three different coarsenesses of the microstructure and a low amount of defects.

Keywords: Cast aluminum alloys, Heat Treatment, Solution Treatment, Artificial Ageing, Tensile Properties, Plastic Deformation, Microstructure, Modeling

INTRODUCTION

Aluminium is a material of special interest due to its high strength to weight ratio. Further beneficial properties are that it is easy to recycle, it is corrosion resistant and has a high electrical and thermal conductivity. Less beneficial is its low strength at high temperature and low stiffness. Aluminium is used in five major areas; building and construction, containers and packaging, transportation, electrical conductors, machinery and equipment (I J Polmear, 2000).

Aluminium alloys are designated as wrought and cast alloys. The Al-Si alloys are the largest group of cast alloys due to their excellent cast ability. Addition of Si increases the fluidity and decreases the solidification shrinkage, resulting in an increase in castability.

A further advantage is that Si can be added without increasing the density of the alloy. Si increases the strength and stiffness, but reduces the ductility. Commercial Al-Si casting alloys have Si concentrations in the range of 5 to 23 wt%. Three different microstructures form depending on the Si concentration, i.e. the alloy can be hypoeutectic, eutectic or hypereutectic, see

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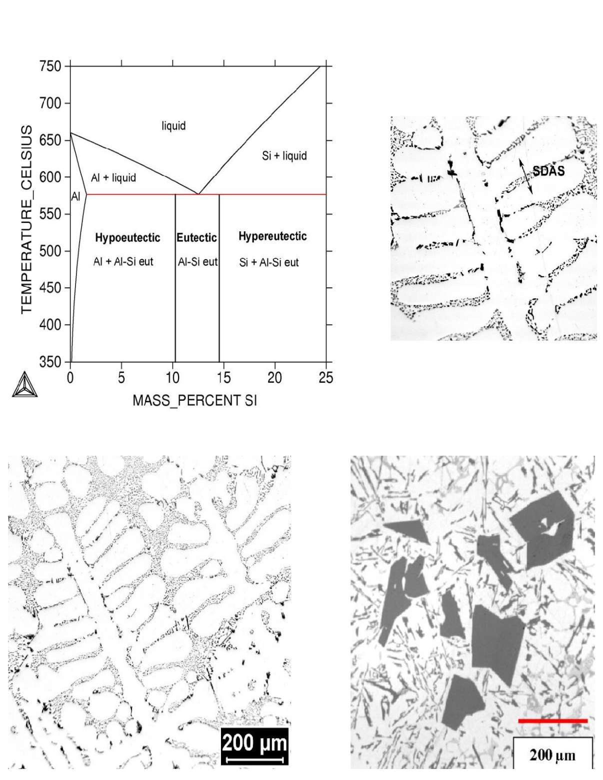
Figure 1a. The microstructure of the hypoeutectic alloys consists of α-Al dendrites which solidify first followed by the Al-Si eutectic, see Figure 1c. The distance between the secondary dendrite arms, SDAS, in Figure 1b is related to the local solidification time. Primary Si particles form first in hypereutectic alloys followed by the Al-Si eutectic, see Figure 1d. The Si concentration of alloys used in the automotive industry often ranges between 5 and 10 wt% and they are frequently used in applications such as engine blocks, cylinder heads and wheels. Hypereutectic alloys are used when increased wear properties are needed (B Thundal, 1991). Typical applications are cylinder liners, pistons and piston rings.

Figure 1: a) Al-Si Phase Diagram. b) Illustration of SDAS. Microstructures of c) Hypoeutectic and d) Hypereutectic Alloys



HISTORY

Compared to other metals such as iron, copper, tin, etcaluminium is a relatively new material. Pure aluminium was first isolated around 1825. In 1845 a piece large enough to study its properties was extracted and it was concluded that aluminium is ductile and non magnetic (B Thundal, 1991). The first factory for aluminium production was built in 1854 close to Paris. At this time aluminium was still produced by chemical means and was more expensive than gold (I J Polmear, 2000). In 1886 Hall and Héroult independently produced aluminium by electrolysis of alumina, Al_2O_3, dissolved in molten cryolite (B Thundal, 1991). The production of high purity alumina from bauxite around 1890 was the last step in reaching a cost effective production of aluminium (B Thundal, 1991). The electrolysis is however very energy intensive and the aluminium production plants (smelters) are placed in regions where electricity is cheap. Aluminium products have a long life and can be recycled using only 5% of the production energy.

In 1906 Dr. Alfred Wilm investigated the possibility to increase the strength of aluminium alloys (E Hornbogen and J Light Met, 2001; I J Polmear, 2004). He knew that the strength of steel could be increased by a high temperature treatment followed by a fast quench. Wilm applied the same procedure to aluminium alloys. Disappointed he observed a decrease rather than an increase in strength. However, at one occasion, Wilm quenched his samples and then left for the weekend. Returning, he was surprised to find that the strength of the samples had increased. Age hardening was discovered! The alloy

Wilm used contained 3.5-5.5% Cu and less than 1% Mg and Mn and was subsequently named Duralumin. The discovery was followed by a search for other aluminium alloys which age harden.

The Relation Between Microstructure and Tensile Properties

The tensile properties of Al-Si casting alloys can be altered within a large spectrum by the choice of 1) alloy composition, 2) casting process and 3) post solidification treatment.

Alloy Composition

The Si particles have a plate-like morphology in unmodified aluminium alloys, which act as crack initiators and have a negative influence on ductility. The alloy ductility can be improved by changing the morphology of the Si particles towards a more fibrous form. This can be done by using a high cooling rate, by addition of a chemical modifier, by exposing the casting to a high temperature for long periods, or by a combination of these processes. Strontium, Sr, is often used as a chemical modifier and small additions of 50-300 ppm are often made. The Sr concentration needed to obtain a fibrous morphology depends on the purity of the melt; Mg for example negates some of the modifying effect. The fibrous morphology obtained through Sr modification is much easier to fragment and spheroidize during solution treatment and the solution treatment time can be shortened.

Cu, and Mg are added to increase the strength of the alloy, but this also lead to a reduction in ductility. The strength and ductility obtained are affected by factors such as if the Cu and Mg are present as coarse phases after solidification, as atoms in solid solution, as GP zones formed at room temperature, or as precipitates formed during artificial ageing. The coarse phases which may form during solidification are the Al_2Cu phase and the Q-$Al_5Mg_8Si_6Cu_2$ phase in Al-Si-Cu-Mg alloys, while the π-$Al_8Mg_3FeSi_6$ phase and the Mg_2Si phase form in Al-Si-Mg alloys. These coarse phases do not contribute to strength and their degree of influence on ductility depends on their distribution and size relative to the Si particles. The strength increase obtained in the as-cast condition arises from atoms in solid solution and from GP zones which form at room temperature. The highest strength contribution is obtained when Cu and Mg are present as small precipitates after a heat treatment, but a reduction in ductility also results. Additions of Cu and Mg also leads to the formation of bands of coarse Si particles and an increased risk for shrinkage porosity due to an increased solidification interval, which may decrease the elongation to fracture.

Casting Process

The casting process determines both the solidification rate and the defect content. Typical defects found in Al-Si casting alloys are oxides and pores. The defect content depends on the cleanliness of the melt and how it is introduced into the mould, while the solidification ratedepends on the geometry of the component (wall thickness and mass centre) and the ability to remove heat from the casting (mould material, chill, water cooling).

The solidification rate determines the coarseness of the microstructure including the fraction, size and distribution of intermetallic phases and the segregation profiles of solute in the α-Al phase. Large and brittle intermetallic phases form during a slow solidification, which may initiate or link fracture, decreasing elongation to fracture. Additionally, the defect size such as pore size, is also controlled to some extent by the solidification rate. The influence of defects on the elongation to fracture depends on their size, shape, distribution and fraction.

Post Solidification Treatment

The post solidification treatment of interest for Al-Si casting alloys is heat treatment, while both cold working and heat treatment are of interest for wrought alloys. The main reason for doing a heat treatment is to obtain an increase in strength. Different heat treatment processes are available depending on the casting process and the desired properties of the alloy. A T6 heat treatment; consisting of a solution treatment, a quench and an artificial ageing, is often used for gravity cast components to achieve an increase in strength. The T6 heat treatment, which is the focus of this thesis, is discussed in more detail in the next section. The T6 heat treatment can not be used for HPDC components, as they can not be solution treated at a high temperature due to blistering. A T5 heat treatment, consisting of a fast cooling after solidification together with an artificial ageing is instead used to improve the strength. A short solution treatment prior to artificial ageing has however been shown to be successful in increasing the yield strength above that of a T5 heat treatment without occurrence of blisters.

THE T6 HEAT TREATMENT

A T6 heat treatment consists of the following stages:

1. Solution treatment at a high temperature, close to the eutectic temperature of the alloy. The purpose of the solution treatment is to:
   a. dissolve Cu- and Mg- rich particles formed during solidification
   b. homogenize the alloying elements
   c. spheroidize the eutectic Si particles.
2. Quenching, usually to room temperature, to obtain a supersaturated solid solution of solute atoms and vacancies.
3. Age hardening, to cause precipitation from the supersaturated solid solution, either at room temperature (natural ageing) or at an elevated temperature (artificial ageing).

The T6 heat treatment is illustrated in Figure 2 for an Al-Si-Cu alloy as an example. The evolution of the microstructure is shown; from 1) atoms in solid solution at the solution treatment temperature, through 2) a supersaturated solid solution at room temperature after quench, to 3) precipitates formed at the artificial ageing temperature.

Solution Treatment

The time needed for solution treatment depends on the as-cast microstructure, i.e. the size, distribution and type of intermetallic phases and the morphology of the Si particles, as well as on the temperature used. A high solution treatment temperature gives a faster dissolution, homogenization and spheroidization, and the solubility of alloying elements is higher, which will result in a higher yield strength after artificial ageing. The temperature that can be used is limited by incipient melting of phases formed from the last solidified melt that is rich in solute elements due to segregation. Localised melting results in distortion and substantially reduced mechanical properties. Cast Al-Si-Mg alloys can be solution treated at 540-550°C, while alloys containing Cu must be solution treated at a lower temperature due to the risk of local melting of Cu-containing phases. According to Samuel Cu-containing phases start to melt at 519°C in an A319 alloy with low Mg concentration, while melting starts at 505°C in an A319 alloy with 0.5 wt.% Mg, due to the presence of the Q phase. The exact temperature that can be used without localized melting depends on the solidification rate and the heating rate to the solution treatment temperature. A two step solution treatment can be used for Cu-containing alloys to increase strength and ductility. The alloy is then first solution treated at a low temperature to dissolve Cu-rich phases and then at a higher temperature to increase the speed of homogenisation and spheroidization.

Not all phases will dissolve during a solution treatment. The Q phase is reported to be stable or to dissolve very slowly for alloys having a high Cu concentration (3.5-4.4 wt %) and various Mg concentrations when solution treated at 500°C. The π-Fe phase transforms into the β-Fe phase and Mg in solid solution when the Mg concentration is low (0.3-0.4 wt %). The transformation does not take place or may be reversed when the Mg concentration is high (0.6-0.7 wt %).

Quench

The objective of quenching is to suppress precipitation upon cooling of the casting from the high solution treatment temperature to room temperature. If the quench rate is sufficiently high a high concentration of solute in solid solution and of vacancies is retained. On the other hand if the cooling is too slow, particles precipitate heterogeneously at grain boundaries or dislocations, which results in a reduction in super saturation of solute and concomitantly a lower maximum yield strength after ageing. In Al-Si casting alloys; Si may diffuse from the matrix to eutectic Si particles and Mg2Si phases may form on the eutectic Si particles or in the matrix, reducing the super saturation of Mg and Si in the matrix.

The quench rate is especially critical in the temperature range between 450°C and 200°C for most Al-Si casting alloys where precipitates form rapidly due to a high level of super saturation and a high diffusion rate. At higher temperatures the super saturation is too low and at lower temperatures the diffusion rate is too low for precipitation to be critical. 4°C/s is a limiting quench rate above which the yield strength increases slowly with further increase in quench rate.

Artificial Ageing

A general and simplified precipitation sequence can be described as follows. After solution treatment and quench the matrix has a high super saturation of solute atoms and vacancies. Clusters enriched in Si and Mg atoms form rapidly from the supersaturated matrix and evolve into GP zones. Metastable coherent or semi-coherent precipitates form either from the GP zones or from the supersaturated matrix when the GP zones have dissolved. The precipitates grow by diffusion of atoms from the supersaturated solid solution to the precipitates. The precipitates continue to grow in accordance with Ostwald ripening when the super saturation is lost. The length of each step in the sequence depends on the thermal history, the alloy composition and the artificial ageing temperature.

In Al-Si-Mg alloys separate clusters of Mg and Si atoms form initially, which develop into co-clusters. GP zones form from the co-clusters, which elongate and transform into the β” -Mg5Si6 phase, which is the phase having the greatest strength contribution. Upon over ageing some of the β” phases transform into the rod-like β’ phase. The Mg:Si ratio increases through the precipitation sequence, which makes the super saturation of Si an important parameter as it influences the fraction of precipitates formed during initial ageing.

The precipitation sequence for Al-Si-Cu-Mg alloys is similar, but more complex, as the Q” phase and the θ’ phase may also form. Cu can increase the fraction of the β” phase formed, but it can also form the Q” phase, which has a lower strength contribution compared to the β” phase. The β” phase is therefore preferred, rather than the Q” phase. It is however not clearly stated when the Q” phase forms at the expense of the β” phase in cast alloys. For wrought alloys it has been shown that the fraction of the Q” phase increases with natural ageing and artificial ageing time and temperature.

The precipitation sequence in Al-Si-Cu alloys is influenced by the high density of dislocations formed during quenching due to the difference in thermal expansion between the Si particles and the α-Al matrix. Fine and evenly dispersed θ” phases form in the centre of the dendrites, while coarse θ’ phases form on the dislocations close to the Si particles. The coarse θ’ phases have a negligible strength contribution and can be seen as a loss of Cu atoms that could have increased the fraction of θ” phases. These three alloy groups show different age hardening response, which is the increase in yield strength on artificial ageing compared to the yield strength in the as-quenched or natural aged condition. The age hardening response depends on the fraction, size, distribution and coherency of precipitates formed. Al-Si-Cu-Mg alloys and Al-Si-Mg alloys generally have a high age hardening response, while Al-Si-Cu alloys have a slow and low age hardening response.

SUMMARY OF RESULTS AND DISCUSSION

Influence of solidification rate and alloying elements on the as-cast micro- structure (supplements ii and iii)

The solidification rate and the alloying elements determine the as-cast microstructure, i.e. 1) the coarseness of the microstructure, 2) the segregation profiles of alloying elements in the α-Al matrix and 3) the type, size and distribution of the intermetallic particles.

Coarseness of the Microstructure

SDAS is often used as a measure of the coarseness of the microstructure. The relationship between the solidification rate and the coarseness of the microstructure, SDAS, is well studied in the literature, and is not the focus of the present study. The average measured SDAS and standard deviations within brackets for the investigated alloys are presented in Table 2. A small influence of the amount of alloying elements on the SDAS can be observed. The solute lean Al-Si-Mg alloy has a slightly higher SDAS compared to the alloy containing 3 wt% Cu, which is in agreement with data reported by Shabestari.

CONCLUDING REMARKS AND FUTURE WORK

Concluding Remarks

The aim of creating a model for the yield strength of heat treated Al-Si alloys was reached for Al-Si-Mg alloys. The development of a model for the plastic part of the stress-strain curve has been initiated with a comparison of different models available in the literature. During the work towards modelling of the tensile properties of heat treated alloys new knowledge has been gained concerning the microstructure formation during solidification and its evolution during heat treatment, and the relation between microstructure and strength.

Table 1: As-cast Parameters of the Microstructure; SDAS and Length of the Largest Intermetallic Particles. Standard Deviations Within Brackets.

Alloy	sol. rate [mm/s]	SDAS [µm]	Length of particles[a]			
			Al_2Cu [µm]	Fe rich [µm]	Si [µm]	Q [µm]
Al-Si-Mg	0.03	51 (7)		31 (20)	22 (7)	
	0.3	28 (3)		14 (9)	8 (2)	
	3	10 (1)		3 (1)	2 (1)	
Al-Si-Cu	0.03	50 (6)	193 (86)	109 (29)	44 (11)	
	0.3	25 (4)	57 (20)	41 (16)	12 (3)	
	3	10 (2)	8 (3)	-	2 (1)	
Al-Si-Cu-Mg	0.03	49 (7)	107 (41)	87 (28)	28 (8)	40 (24)
	0.3	24 (3)	30 (10)	24 (11)	10 (2)	11 (8)
	3	9 (1)			3 (1)	

[a]The phases in the Al-Si-Cu-Mg alloy having the finest microstructure were too small and close to each other to distinguish the different phases.

Solidification (Supplements I, II & III)

The π-Fe phase was the main Mg containing phase formed in the Al-Si-Mg alloy. The fraction of the π-Fe phase was observed to increase, while the fraction of the β-Fe phase decreased with increasing solidification rate. The Al2Cu phase was the main Cu-containing phase in the Al-Si-Cu and Al-Si-Cu-Mg alloys. The Q phase formed in addition to the Al2Cu phase in the Al-Si-Cu-Mg alloy. These observations are in agreement with the literature.

The distance between the Mg/Cu rich phases increases faster than the SDAS increases with decreasing solidification rate. There are however indications that these observations may be limited to directional solidified samples, while the distribution of Mg/Cu rich phases is more uniform for equiaxed solidified samples.

The segregation profile of Mg is influenced by back diffusion giving a higher Mg concentration in the centre of the dendrite arms for a slow solidification as diffusion is allowed to take place over a longer time. The segregation profiles for Cu are only weakly dependent on the solidification rate due to the slower diffusivity of Cu compared to Mg in the α-Al phase.

Solution Treatment (Supplements I, II & III)

ThermoCalc can predict the solubility limit of Si at the solution treatment temperature well. Furthermore the stability of the phases in the Al-Si-Mg and Al-Si-Cu alloy is correctly predicted. For the Al-Si-Cu-Mg alloy a too high solubility of Mg in the α-Al phase at 495°C was predicted.

The time needed for dissolution and homogenisation is strongly dependent on the coarseness of the microstructure. As an example; 10-30 min. at 495°C was sufficient to achieve complete dissolution and homogenisation for the finest microstructure of the Cu-containing alloys, while more than 10 h was needed for the coarsest microstructure. The solution treatment process was shown to be diffusion controlled using a dimensionless diffusion time. The diffusivity of the alloying elements was not influenced by the presence of other alloying elements.

Artificial Ageing (Supplements I, IV & V)

There is a good agreement between different literature investigations regarding the yield strength of artificially aged Al-Si-Mg alloys. The measurements conducted in this work are also in good agreement with data sets from the literature. The situation is more complex for Al-Si-Cu-Mg alloys as different precipitates form and there is a large scatter in the data found in the literature. Different results are obtained from what seem to be identical investigations. Measurements made in this work does not contribute to any clarification regarding this issue, but the results are in line with data from the literature and existing ideas for wrought alloys. The KM strain hardening theory provides information regarding the coherency of the precipitates. From these studies it is indicated that the precipitates of the Al-Si-Mg alloy remain coherent on initial overageing, while some of the precipitates of the Al-Si-Cu-Mg alloy lose coherency directly on overageing. For the Al-Si-Cu alloy there is already a mixture of coherent and non-coherent precipitates on underageing, changing into only non-coherent precipitates on overageing.

The elongation to fracture increases directly after the peak yield strength for the Al-Si-Cu-Mg alloy having the finest microstructure, while it continue to decrease on initial overageing for the Al-Si-Mg alloy. This is suggested to be due to formation of Si precipitates in the matrix. The coarser microstructures of the Al-Si-Cu-Mg alloy showed a low elongation to fracture around 1% for ageing times corresponding to a high yield strength. The high yield strength in combination with larger defects in the coarser microstructures is thought to be the reason for the low elongation to fracture.

Modelling (Supplements II, III, V and VI)

A solution treatment model that can handle the three alloys investigated has been developed. The solubility limit of Mg at the solution treatment temperature was adjusted for the Al-Si-Cu-Mg alloy due to the presence of the stable Q phase. A reduced diffusion distance between the Al2Cu phases only including the primary ?-Al phase was introduced due to a reported high diffusivity of Cu in the Al-Si eutectic.

A model for the yield strength of heat treated Al-Si-Mg casting alloys has been developed which can handle Mg concentrations in the range 0.2 wt% to 0.6 wt% and temperatures between 150°C and 210°C. The model was calibrated to a measured ageing curve, but all parameters have physical relevance. The model predicts the yield strength on overageing well, while the underaged condition is less well described. The model has been implemented in a development version of MAGMAnon-ferrous.

The Hollomon equation can be used to describe the plastic deformation when shearable precipitates are present, while the Ludwigson equation is needed when a supersaturated solid solution is present. The plastic curve of the Al-Si-Cu alloy has a complicated shape which is thought to be caused by non-coherent precipitates and the Hollomon and Ludwigson equations can not describe the shape of the curve very well. The KM model for pure metals can be used to model the plastic deformation for Al-Si-Mg and Al-Si-Cu-Mg alloys. A modified version of the KM model for non-coherent precipitates was used for the Al-Si-Cu alloy which gave a perfect fit to the whole plastic part of the stress-strain curve.

FUTURE WORK

Solidification

The distance between Mg/Cu rich phases is measured and used as an input to the solution treatment model. By investigating the position of the phases in the microstructure by for example grain boundary etching, the distance between the phases could in the future be modelled. The measured distance between the phases should also be compared with equiaxed solidified samples to see if there is a difference. Back diffusion can be included in the model to predict the yield strength for the as-cast condition.

Solution Treatment

The reported high diffusivity of Cu in the Al-Si eutectic should be verified by solution treatment of a eutectic Al-Si-Cu alloy or a binary Al-Cu alloy.

To widen the alloys the model can handle the phase diagram of Al-Si-Cu-Mg alloys need to be studied.

Quench

Quench and natural ageing is not included in the present model. A rapid quench and no natural ageing are assumed. Quench and natural ageing is however known to influence the shape of the ageing curve and should be included in the model.

Artificial Ageing

Experiments are needed to determine which are the parameters that influence the fractions of the β” and Q” phases formed during artificial ageing in order to be able to develop a yield strength model for Al-Si-Cu-Mg alloys.

The modelling of yield strength for the underaged condition of the Al-Si-Mg alloy needs further attention, for example to see if small clusters could be introduced in the model instead of Mg and Si atoms in solid solution.

If mathematical relations between the parameters ($\theta 0$, β, K) in the KM strain hardening theory and the microstructure (solid solution concentration, fraction and radius of precipitates) can be derived for various alloy groups a model for plastic deformation of defect free alloys can be developed. The UTS and elongation to fracture can then be derived using Considère's criterion for necking.

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