J Mater Sci (2008) 43:3577–3581
DOI 10.1007/s10853-008-2565-6

Semisolid processing of near-eutectic and hypereutectic Al–Si–Cu alloys

Yucel Birol

Received: 8 October 2007 / Accepted: 19 February 2008 / Published online: 29 March 2008

Abstract Semisolid processing of near-eutectic and hypereutectic versions of alloy 380 offers to overcome the problems encountered in casting hypereutectic Al–Si alloys and was thus explored in the present work. Experimental near-eutectic and hypereutectic Al–Si–Cu alloys obtained by adding elemental silicon to the 380 alloy were melted and were cooled to within 5 to 15 °C of their liquidus points before they were poured into a permanent mould in order to produce non-dendritic feedstock for thixoforming. This low superheat casting (LSC) process largely replaced α-Al dendrites with relatively smaller α-Al rosettes in all alloys. The slugs machined from the LSC ingots thus obtained were thixoformed after they were heated in situ in the semisolid range, between 568 and 573 °C, for 5 min in a laboratory press. Semisolid soaking sufficed to produce the required globular structure even when some dendritic features were retained in the starting feedstock. The hardness of the thixoformed parts which ranged between 84 and 96 HB have increased to 121–131 HB after the T6 heat treatment, implying a considerable age-hardening potential. The T6 treatment also improved the morphology of the eutectic silicon with potential benefits regarding the ductility of the thixoformed part.

Introduction

Forced to manufacture energy-efficient products due to environmental concerns, refrigeration and air-conditioning industries have to use high-performance compressors in their products. One of the practices to achieve better performance in compressors is to use low-viscosity lubricants, which, however, give thinner oil films between connecting rod and wrist pin allowing an increased metal/metal contact and heavier wear problems. This change in the boundary lubrication conditions in service dictates a material more resistant to wear than the current high-pressure die cast (HPDC) 380 alloy [1]. Hypereutectic Al–Si alloys which offer high wear resistance, low thermal expansion, together with excellent castability and reduced density, are then the material of choice [2–4].

However, the use of these alloys in conventional cast grades has been somewhat restricted owing to their high latent heat and consequent long solidification time resulting in die wear, to the segregation and excessive growth of primary silicon particles and to extensive shrinkage [5, 6]. Thixoforming was recently considered to be a viable alternative in the production of these alloys, since the casting temperature and heat content in thixoforming are reduced, the primary silicon is refined and uniformly distributed and the shrinkage is much less than that of a molten alloy [5–7]. The present work was undertaken to explore the semisolid processing of experimental near-eutectic and hypereutectic Al–Si–Cu alloys obtained by adding elemental silicon to the 380 alloy, currently used in compressor connecting rods.

Experimental

Three experimental Al–Si–Cu alloys (E1, E2 and E3) were obtained by adding 4- to 12-wt% Si to the molten 380 alloy (Table 1). Low Superheat Casting (LSC) was employed to produce the thixoforming feedstock [8]. Alloy ingots were melted in an electric resistance furnace set at 750 °C. The

Y. Birol (✉)
Materials Institute, Marmara Research Center, TUBITAK, Kocaeli, Turkey
e-mail: yucel.birol@mam.gov.tr

Table 1 Chemical composition of alloy 380 and the three experimental alloys obtained by adding elemental silicon to it

Alloy	Si	Fe	Cu	Mn	Mg	Zn	Al
380	8.52	1.099	3.099	0.1712	0.1532	1.001	bal
E1	12.65	1.109	2.715	0.1771	0.1256	0.928	bal
E2	15.41	1.056	3.002	0.1618	0.1468	0.966	bal
E3	19.73	1.044	2.841	0.1578	0.1239	0.970	bal

melts were allowed to cool to the pouring temperatures which were estimated from the solidification scans, so as to limit the superheat of the melt as much as possible, and were finally cast into a 30-mm-diameter and 150-mm-deep permanent mould.

The ingots thus obtained were sectioned into 35-mm-long slugs. A medium-frequency induction coil (9.6 kHz, 50 kW) placed right underneath the die was used to heat these slugs in situ into the semisolid state. The melting scans obtained by DSC were used to calculate the change in liquid–solid fractions with temperature. Temperatures for reheating experiments were then estimated from the latter. Temperature was monitored with a K-type thermocouple inserted in a 3-mm-diameter hole drilled in the centre of the slugs. Measures were taken to achieve rapid heating (150 °C min^{-1}) to avoid undesirable grain growth. The required accuracy and the reproducibility of the heating process are achieved by very precise control of the power input into the induction heating unit.

The thixoforming operation was carried out with a laboratory press which employs a pneumatic cylinder to provide the forming load (5 ton-f max). The maximum speed of the ram was 500 mm/s and the die was preheated to 450 °C. The slugs were then soaked in the semisolid range for 5 min to allow spheroidization of the grains. The thermocouple was withdrawn from the sample just before forming and the slurry was formed into an arbitrary shape part. Three trials were made for each of the thixoforming experiments. A second set of thixoformed samples were heat treated to the T6 temper in an air furnace, by solutionizing at 500 °C for 1 h, followed by forced air cooling before ageing at 175 °C for 8 h.

Samples sectioned from as-cast ingots, reheated slugs, thixoformed and heat-treated parts were prepared with standard metallographic practices. These samples were etched with a 0.5% HF solution before they were examined with optical and scanning electron microscopes. X-Ray Diffraction (XRD) was employed for the identification of intermetallic particles with a Shimadzu XRD 6000 Diffractometer equipped with CuK_{α} radiation. The diffractometer was operated at very low scanning rates (0.1–0.5° per minute) to improve the counting accuracy. The hardness of the thixoformed and heat-treated samples were measured in Brinel (HB) units with a load of 31.25 kg and a 2.5-mm-diameter indenter.

Results and discussion

The parent 380 alloy exhibits a dendritic structure of the α-Al solid solution phase with interdendritic eutectic

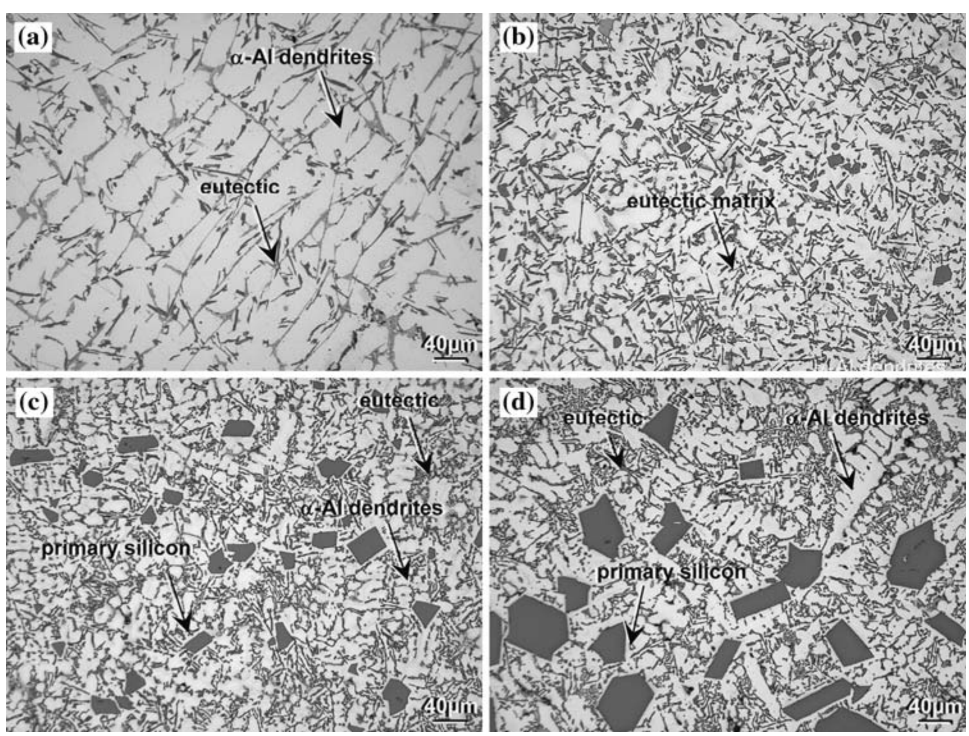

Fig. 1 Microstructure of the as-received 380 alloy ingot (**a**) and the three experimental alloys obtained by adding elemental Si into molten 380: (**b**) E1, (**c**) E2 and (**d**) E3

silicon and occasional intermetallic particles (Fig. 1a). The latter were found by XRD and metallographic analysis to be of the β-Al_5FeSi and $CuAl_2$ variety. The features of alloy E1 are quite similar but reflect its nearly eutectic composition with an almost entirely eutectic matrix (Fig. 1b). Several small primary Si particles show alloy E1 to be slightly hypereutectic. Many more primary Si particles are noted in the next two alloys (Fig. 1c, d). These particles are larger in size and have thus become the predominant feature in alloy E3 which has as much as 20-wt% Si (Fig. 1d). The dendritic features are more frequent in the experimental alloys than in the parent alloy.

The LSC process involves pouring the molten alloys into a permanent mould from the lowest temperatures which allow the entire melt to flow out of the crucible without trouble. After several trials, these temperatures were identified to be 590, 590, 625 and 675 °C for alloys 380, E1, E2 and E3, respectively, i.e. within 5 to 15 °C of the liquidus points measured from their respective solidification scans. The LSC process has evidently produced marked changes in the primary phase morphology of alloy 380 (Fig. 2a). The α-Al dendrites in the as-received ingot are almost entirely replaced by relatively smaller, nearly globular α-Al grains. While not as evident as it is in 380, the dendritic structure in alloys E1 to E3 have also been largely degenerated upon LSC. The fraction of α-Al globules is higher in these alloys after LSC than it is before (Fig. 2b–d).

The solid fraction versus temperature curves, obtained from the melting scans of the LSC ingots thus produced, are shown in Fig. 3. The change in solid fraction with temperature (the slope of the Fs–T curve) between 50 and 70% solid, of interest from the thixoforming stand point, is very similar for all alloys and the curves are only slightly shifted with respect to each other. The temperature to achieve 50% solid is estimated from Fig. 3 to be 567 °C for alloy 380 and increases with increasing Si content, until it is 574 °C in alloy E3. The temperatures at which the thermocouple penetrated into the slug in each case were in reasonably good agreement with the thixoforming temperature range estimated from the Fs–T curves. After several reheating and quenching experiments, the optimum reheating temperatures were determined to be 568, 570,

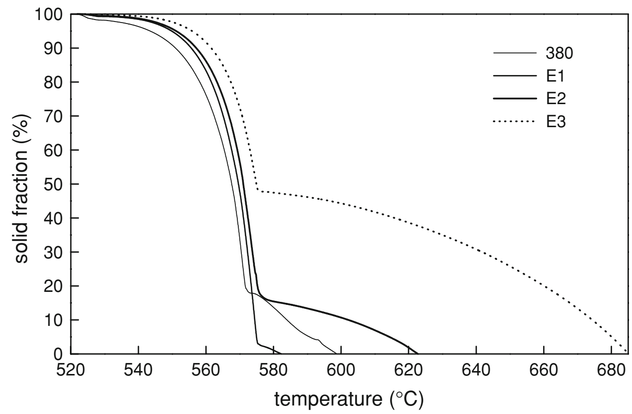

Fig. 3 Change in solid fraction with temperature for alloys 380, E1, E2 and E3

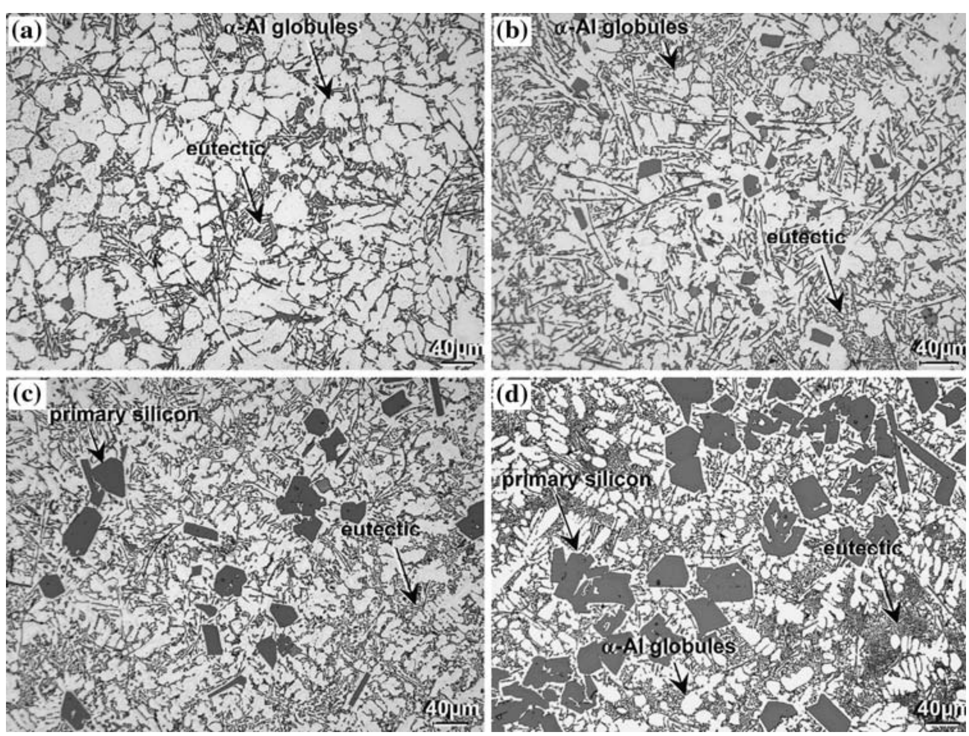

Fig. 2 Microstructure of alloys (**a**) 380, (**b**) E1, (**c**) E2 and (**d**) E3 cast into a permanent mould from 590, 590, 625 and 675 °C, respectively

570 and 573 °C for alloys 380, E1, E2 and E3, respectively.

The slugs machined from the LSC ingots were thixoformed after they were heated in situ at the indicated temperatures for 5 min in a laboratory press. While the LSC microstructures occasionally revealed some dendritic features particularly in alloy E3, the microstructures of the thixoformed parts are dominated by α-Al globules and Si particles sitting in between (Fig. 4). This is taken to imply that semisolid soaking suffices for the present alloys to produce the required globular structure even when some dendritic features are retained in the starting feedstock. α-Al globules and rosettes were found to undergo spherodization and coarsening during reheating before forming which has apparently occurred in the semisolid state.

The hardness of the thixoformed parts range between 84 and 96 HB and have increased to 121–131 HB after the T6 heat treatments, implying a considerable age-hardening potential (Table 2). This hardness level in the thixoformed parts is pleasing considering that the die cast counterparts cannot be heat treated due to extensive blistering linked with porosity. The T6 treatment not only improves the hardness of the thixoformed part but also the morphology of the eutectic silicon with potential benefits regarding the ductility of the thixoformed part (Fig. 5). The eutectic Si needles tend to spheroidize in all thixoformed alloys during solutionizing while the primary Si particles and Fe- and Fe/Cu-based intermetallics resist spheroidization during the solution treatment.

Table 2 Hardness values of the thixoformed parts in the as-formed state and after the T6 treatment

Alloy	Hardness (HB)	
	As-thixoformed	T6 treated
380	83.9 ± 0.5	121.7 ± 0.6
E1	84.8 ± 1.0	121.7 ± 0.6
E2	90.9 ± 0.9	130.3 ± 0.6
E3	95.7 ± 0.9	130.7 ± 1.2

Summary

Three experimental near-eutectic and hypereutectic Al–Si–Cu alloys were obtained by increasing the Si content of alloy 380 by 4 to 12 wt%. These alloys were melted and then cooled to within 5 to 15 °C of their liquidus points before they were poured into a permanent mould in order to produce the non-dendritic feedstock for thixoforming. This process produced marked changes in the primary phase morphology and replaced α-Al dendrites with relatively smaller α-Al rosettes in all alloys. The slugs machined from the LSC ingots thus obtained were thixoformed after they were heated in situ in the semisolid range, between 568 and 573 °C, for 5 min in a laboratory press. Semisolid soaking sufficed to produce the required globular structure even when some dendritic features were retained in the starting feedstock. The hardness of the thixoformed parts which ranged between 84 and 96 HB have increased to

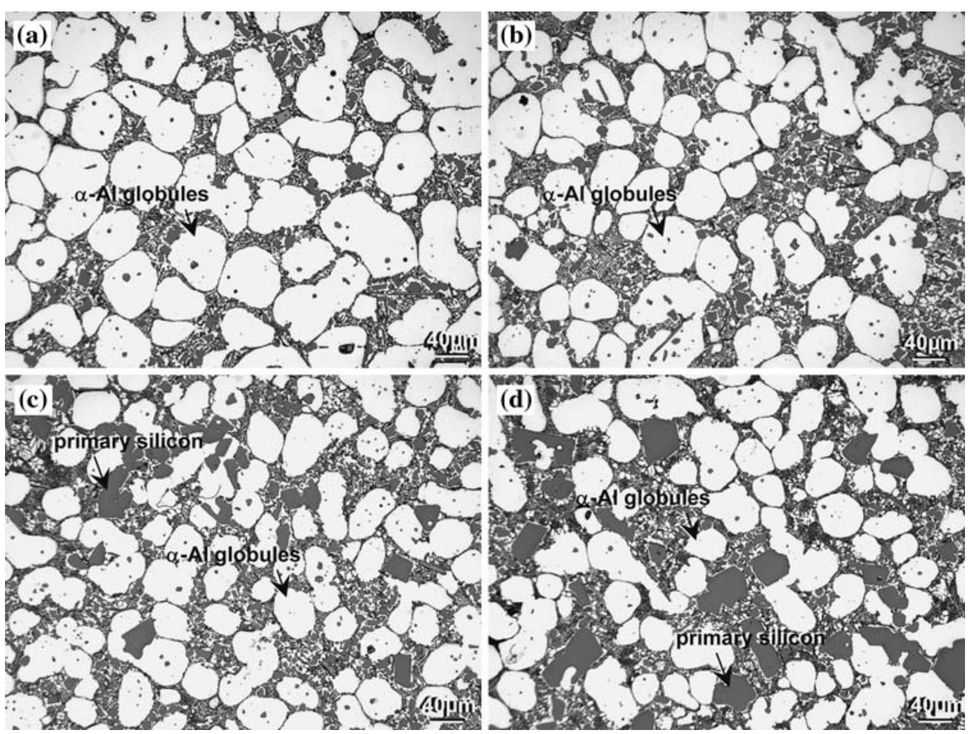

Fig. 4 Microstructure of thixoformed alloys (**a**) 380, (**b**) E1, (**c**) E2 and (**d**) E3

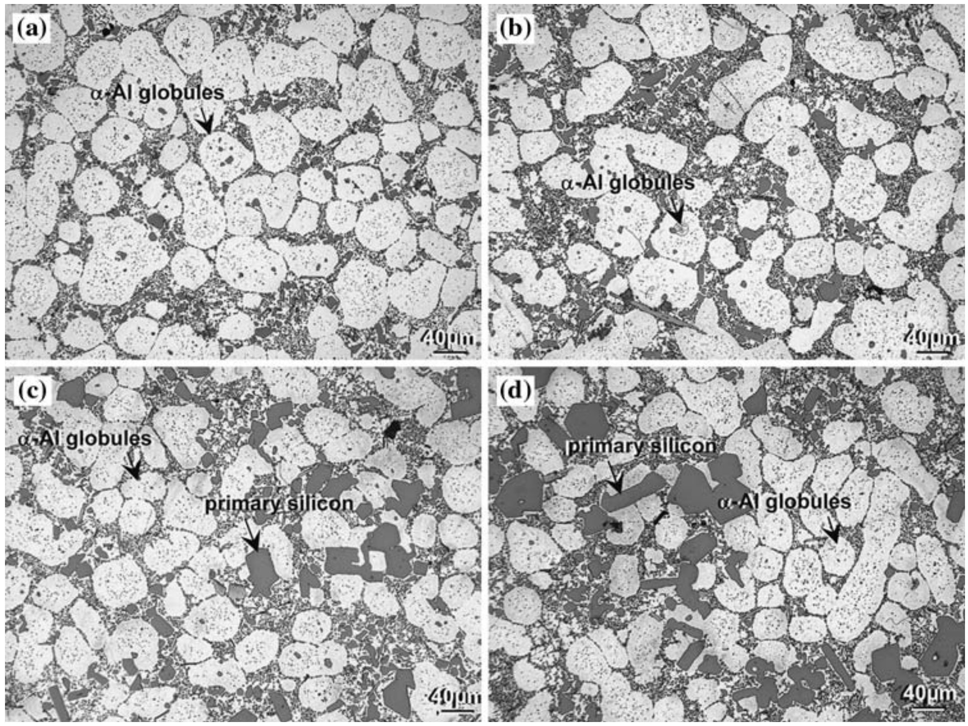

Fig. 5 Microstructure of thixoformed alloys (**a**) 380, (**b**) E1, (**c**) E2 and (**d**) E3 after T6 heat treatment

121–131 HB after the T6 heat treatments, implying a considerable age-hardening potential. The T6 treatment also improved the morphology of the eutectic silicon with potential benefits regarding the ductility of the thixoformed part.

Acknowledgements It is a pleasure to thank O. Çakır and F. Alageyik for their help in the experimental part of this work. The financial support of the State Planning Organization of Turkey is gratefully acknowledged.

References

1. Birol Y, Birol F (2007) In: Cueto E, Chinesta F (eds) 10th ESAFORM conference on material forming. AIP, p 1167
2. Pratt GC (1973) Int Mater Rev 18:1
3. Rohatgi PK, Asthana R, Das S (1986) Int Met Rev 131:115
4. Lasa L, Rodriguez-Ibabe JM (2002) Scr Mater 46:477
5. Kapranos P, Kirkwood DH, Atkinson HV, Rheinlander JT, Bentzen JJ, Toft PT, Debel CP, Laslaz G, Maenner L, Blais S, Rodriguez-Ibabe JM, Lasa L, Giordano P, Chiarmetta G, Giese A (2003) J Mater Process Technol 135:271
6. Midson S, Keist J, Svare J (2002) SAE 2002 World Congress, Detroit, Michigan, March 4–7, 2002, 2002-01-394
7. Ward PJ, Atkinson HV, Anderson PRG, Elias LG, Garcia B, Kahlen L, Rodriguez-Ibabe J-M (1996) Acta Mater 44:1717
8. Birol Y (2007) Int J Mater Res 98:1019