

Structural and mechanical properties of Al–Si alloys obtained by fast cooling of a levitated melt

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Abstract

Data on the structure and mechanical properties of cast Al–Si alloys in a wide compositional range from hypo- to high hyper-eutectic composition are scarce. These properties depend on many factors during solidification of the alloys. In the present work, samples were obtained by rapid cooling of levitated melts of various compositions from 11.5 to 35 wt.% Si. The measurements revealed linear concentration dependences of density and Young's modulus. The average temperature coefficient of Young's modulus in the range from room temperature to 500 °C and the yield point for bending both had maxima at about 20 wt.% Si. The hysteresis of the temperature dependence of Young's modulus had a minimum at about 20 wt.% Si as well. Changing Young's modulus temperature coefficient and Young's modulus hysteresis as a function of the Si content are connected with the creation of the Guinier–Preston zones. Values of the yield point are explained by the plasticity of components of the eutectic structure, primary crystals and grain boundaries. The extrema of the concentration dependences of the mechanical properties occurred for the fine-grained structure arisen from coupled eutectic-like growth. Solidification at other conditions led to formation of primary crystals of α solid solution or primary Si crystals.

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1. Introduction

Al–Si alloys are widely used in different fields of industry. Various additives are usually used to modify industrial alloys. Recently, much attention has been given to unmodified cast alloys, especially to hypereutectic Al–Si alloys. At the same time, the structure and mechanical properties of hypereutectic unmodified cast alloys has been studied mainly for Si content up to 19 at.%. It is only known that increasing the Si content results in an increase of the strength of hypoeutectic alloys and a decrease of the strength of hypereutectic alloys [1,2].

In addition to chemical composition, the structural and mechanical properties of alloys depend on many factors that

act during solidification. Important factors are the structure of the melt, the crystallization rate, and the temperature gradient at the liquid–solid interface [1–5]. As a rule these factors are varied simultaneously, giving rise to contradictory information on the structure and mechanical properties of Al–Si alloys. Thus, for example, the yield stress $\sigma_{0.2}$ was published to increase [2,4] or decrease [1] with increasing content of Si. In order to investigate the influence of the Si content on structure and mechanical properties, it is necessary to prevent contamination by impurities from the crucible and the environment, to maintain constant the superheating of the melt, to have a constant and rather high cooling rate, and effective mixing of the molten alloy.

The aim of this work was to determine the influence of Si content on the structure and mechanical properties of near eutectic and hypereutectic Al–Si alloys with high Si content obtained by fast cooling.

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We used electromagnetic levitation of the melt for improved mixing and stirring of the liquid metal as well as to avoid contamination by impurities from the crucible. The melt was cast in the metal mould. It was anticipated that difference in the structure and properties of castings obtained at constant rather rapid cooling of the melt would be determined by the composition of the alloys only.

The effect of mechanical vibration and electromagnetic stirring on the structure of solidified Al–Si alloys was studied in [6–8]. It was shown that mechanical vibration at varying frequencies between 15 and 41.7 Hz had reduced the shrinkage pipes and produced grain refinement but caused coarsening in the eutectic silicon and of primary silicon of Al–12.5 wt% Si [6]. In [7,8] the possibility to refine Si particles by crashing them during cavitation of bubbles in the melt under electromagnetic field was investigated. In our work the solidification of the liquid took place in the mould outside the region of electromagnetic field so there were no cavitation during crystallization. Earlier levitation melting was used in [9] to study the effect of undercooling on solidification of Al–50 wt% Si hypereutectic composition.

2. Experimental procedure

In the present work, 99.999% pure Al and semiconductor grade n-type Si of 2000 Ω cm specific resistivity were used. Levitation of Al–Si alloys was achieved by applying an alternating inhomogeneous magnetic field produced by a levitation coil. The levitation coil was connected to a RF-generator of 300–600 kHz, and 10 kW power was used to produce the levitation field. The coil and the sample were housed in a vacuum chamber made of stainless steel and equipped with windows to allow visual observation by a video camera and temperature measurement by a pyrometer. The melting was carried out at temperatures from 830 to 1040 °C. The liquid metal was cast into a massive Cu mould by decreasing the inductor voltage. Castings had the shape of 4 mm diameter rod 50 mm in length. The melting and the casting were carried out in a He atmosphere of 1×10^5 Pa pressure. The alloys of Al–Si system of 11.5, 20, 25, 30, 35 wt.% Si were prepared.

Sections were cut with a diamond saw and polished for microstructure characterization, which was performed on alloys as-cast and after heating to 500 °C during measurement of Young's modulus as a function of temperature. Samples for measurement of density and mechanical properties were cut by an electric spark technique.

The density ρ was measured by hydrostatic weighting. The error of measurements of ρ did not exceed 0.5%. Young's modulus E from room temperature to 500 °C was determined on rod samples of 10–30 mm in length by a resonant electrostatic method [10] using longitudinal oscillations at a frequency of 100–140 kHz. The error of measurements of E with an account of the dispersion for different samples was about 1%. The precision of measurements of temperature dependence of E for a separate sample was about 10^{-3} %. PMT

hardness tester with indentation loads of 5 kg was used to measure Vickers hardness, HV. Deformation tests by three-point bending were performed with an Instron 1341 testing machine at the constant sag rate of about 0.5 μ m/s, the load measurements being made at every second. The dimensions of samples for mechanical tests were 1.6 mm \times 2.5 mm \times 20 mm.

3. Results

Optical microscopy revealed a fine-grained structure for samples prepared by quenching of liquid metal. Fig. 1 shows the optical micrographs of Al–Si alloys obtained by casting of levitated melts for (a) 11.5, (b) 20, (c) 25, (d) 30 and (e) 35 wt.% Si. In the hypoeutectic alloy (Fig. 1a) there is eutectic structure with primary crystals of Si solid solution in Al. The 20 wt.% Si alloy (Fig. 1b) has an anomalous finest-grain eutectic structure. There are star-shaped primary Si crystals on the background of eutectic structure for the 25 wt.% Si composition (Fig. 1c). The concentration and size of primary Si crystals increase with increasing Si content (30 and 35 wt.% Si).

Fig. 2 shows that the density of the alloys decreases almost linearly with Si content.

Fig. 3 shows that Young's modulus E increases with increasing Si content. It is noteworthy that alloys prepared by quenching of the levitated melt had higher E values than those obtained with conventional casting [5].

The Vickers hardness was measured only for samples with Si content up to 15 wt.%. The Vickers hardness of transverse sections was slightly higher for 15 wt.% Si alloy (HV = 440 ± 10 MPa) than for the sections of 11.5 wt.% Si alloy (HV = 430 ± 20 MPa).

Fig. 4 shows the temperature dependence of Young's modulus of Al–Si alloys with 11.5, 20, 25 and 30 wt.% Si as well as for Al–11.5 wt.% Si with added 0.01 wt.% Sr. Note that there is temperature hysteresis of Young's modulus for all alloys quenched from the levitated state except for the Al–11.5 wt.% Si–0.01 wt.% Sr alloy. The hysteresis depended on the alloy composition.

Fig. 5 shows the results of the bending test. Note that the load P corresponding to constant sag S of bending deformation had a maximum near 20 wt% Si.

4. Discussion

Increasing the Si content in Al–Si alloys resulted in decreasing of ρ and increasing of E . These dependencies were close to linear, indicating that quenching of the melt with various Si content took place under the same conditions. The linearity of concentration dependence of ρ suggests that the samples had few, if any, pores. The small deviation from linearity of the concentration dependence of E at room temperature can be attributed both to less density of structure defects

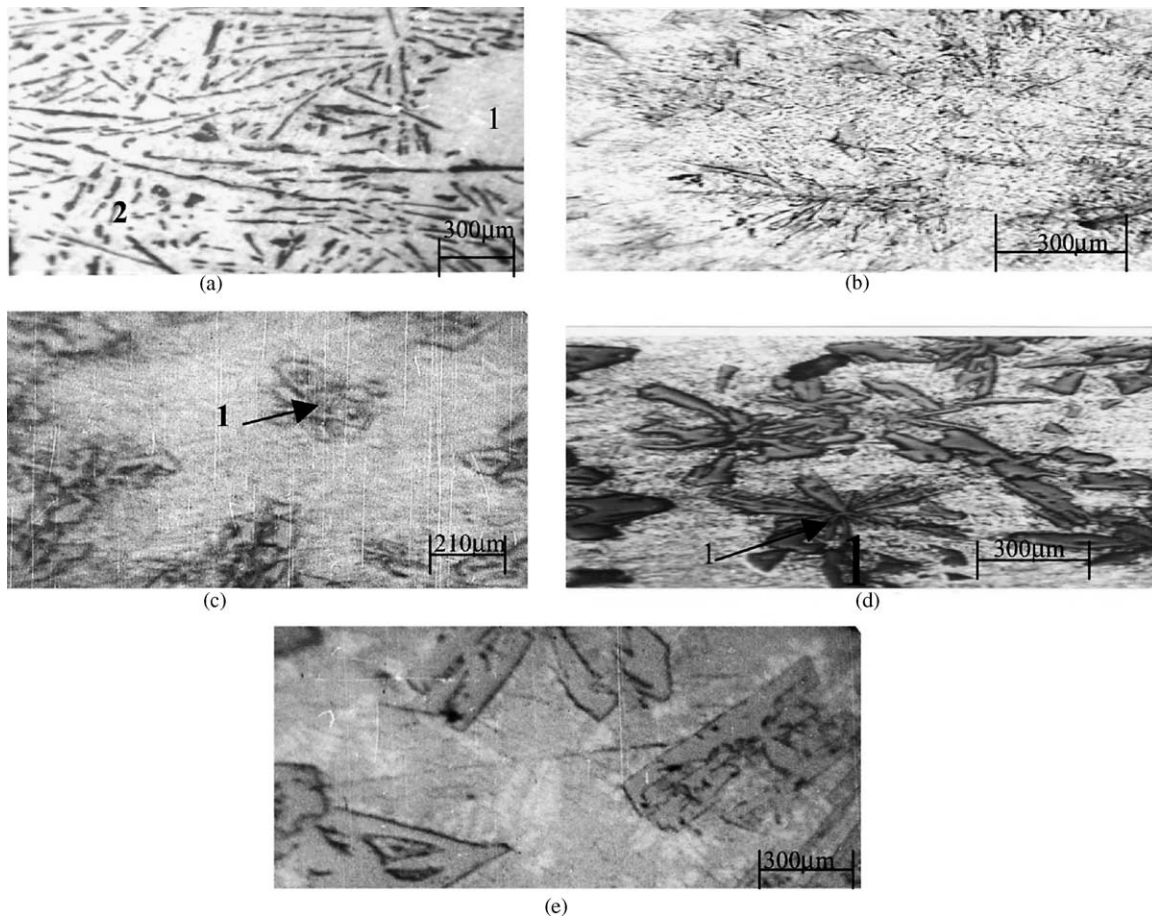


Fig. 1. Microstructure of Al-Si alloys vs. Si content: (a) 11.5 wt.% Si, 1 is the primary crystal of α solid solution, 2 is the eutectic structure with needle like Si component of black colour. (b) 20 wt.% Si. (c) 25 wt.% Si, 1 is star like primary Si crystal. (d) 30 wt.% Si, 1 is star like Si primary crystal. (e) 35 wt.% Si.

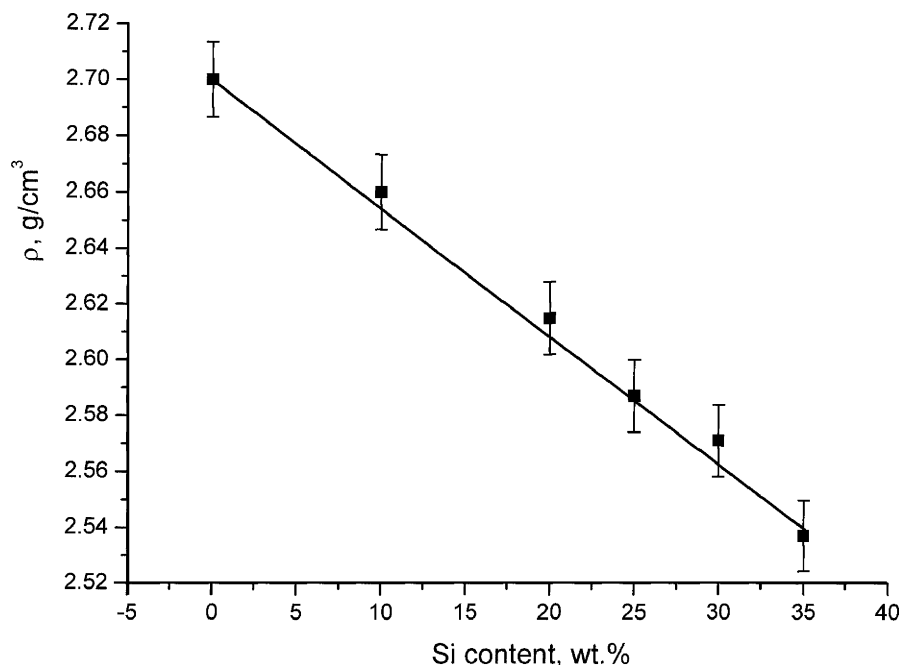


Fig. 2. Density of Al-Si alloys as a function of Si content (in wt.%).

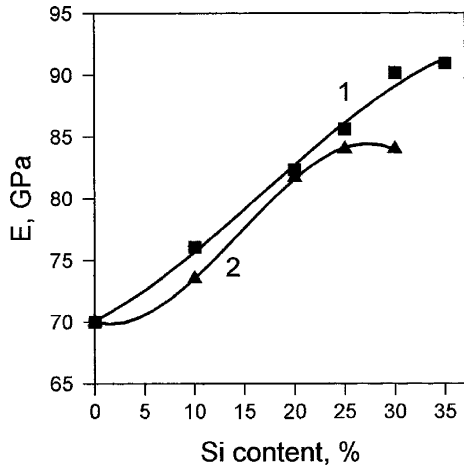


Fig. 3. Young's modulus of Al-Si alloys vs. Si content at room temperature: (1) present results, (2) the results for conventional casting into a mould [1].

(dislocations, pores) and to smaller amount of impurities in comparison with ingots obtained by conventional casting to mould. In conventionally cast alloys this dependence was nonlinear [1], as shown in Fig. 3.

The Si content influences both the character of the temperature dependence of E and the change of E at room temperature after heating and cooling of the samples with various Si contents (Fig. 4). Fig. 6 shows the concentration dependence of the average temperature coefficient of Young's modulus, $\Delta E/E\Delta T$, for cooling in the range from 500 to 20 °C obtained using the data of Fig. 4. Fig. 7 shows the hysteresis in Young's modulus, $\Delta E_{20}/E_{20}$, after heating to 500 °C and cooling. The extrema for both dependences occur at about 20 wt.% Si. This behavior can be attributed to different changes in the microstructure during heating of samples of different composition.

Data of Fig. 5 enable us to determine the yield point for bending deformation depending on the composition of the

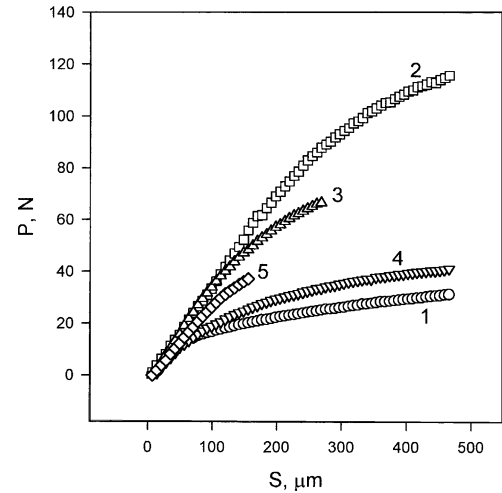


Fig. 5. The sag S dependence on the load P from three-point bending tests of Al-Si alloys: 1—11.5 wt.% Si, 2—20 wt.% Si, 3—25 wt.% Si, 4—30 wt.% Si, 5—35 wt.% Si. Sag rate was about 0.5 μm/s, the P measurements being made at every second.

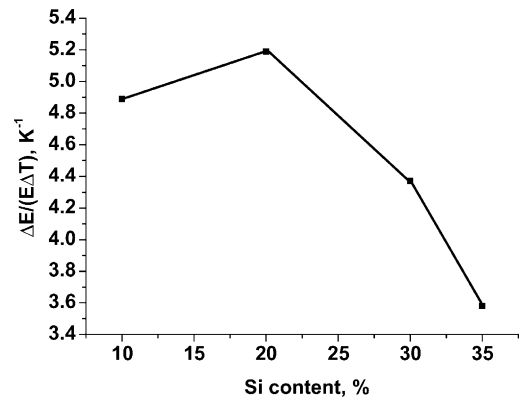


Fig. 6. Concentration dependence of the average temperature coefficient of Young's modulus for cooling in the range from 500 to 20 °C.

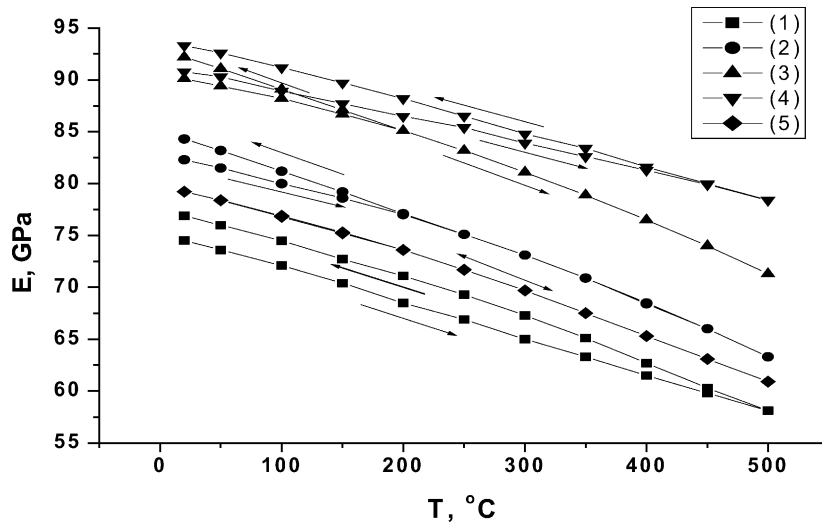


Fig. 4. Young's modulus of Al-Si alloys vs. temperature for various Si contents: 1—11.5 wt.%, 2—20 wt.%, 3—30 wt.%, 4—35 wt.% Si; 5—Al-11.5 wt.% Si-0.01% wt.% Sr.

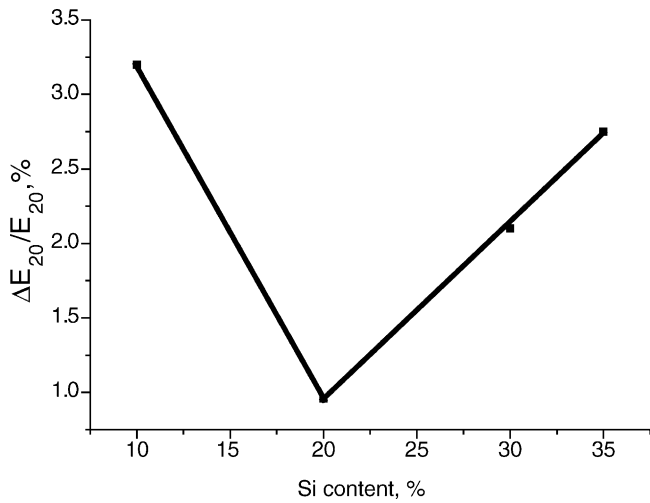


Fig. 7. The change in Young's modulus caused by heating to 500 °C and cooling back to room temperature.

alloy. The yield point P was determined by the intersection of tangents to the *P*–*S* curves for linear (elastic) and nonlinear (plastic) ranges. As shown in Fig. 8a, the load *P* corresponding to the yield point also exhibits a maximum at about 20 wt.% Si. The Si content corresponding to maximum value of the yield point is higher than for the eutectic point of Al–Si alloy (about 11.7%) that is seen from Fig. 8b, where high temperature part of the phase diagram of the alloy is shown.

The mechanical behavior reported above is related to the microstructure, which can be understood with the help of the diagram of microstructure of Al–Si alloys vs. composition and solidification rate, as determined experimentally in [11] for a Si content up to 17%. Fig. 9 shows this diagram with the upper border of the region of coupled growth, the region II, extrapolated by us from 17 to 35 wt.% Si. Note that there is a wide range of Si concentrations and solidification rates giving coupled growth of a fine fiber eutectic-like structure without any primary crystals. The solidification rate was estimated to be between 10³ and 10⁴ K/s. In [12,13] the

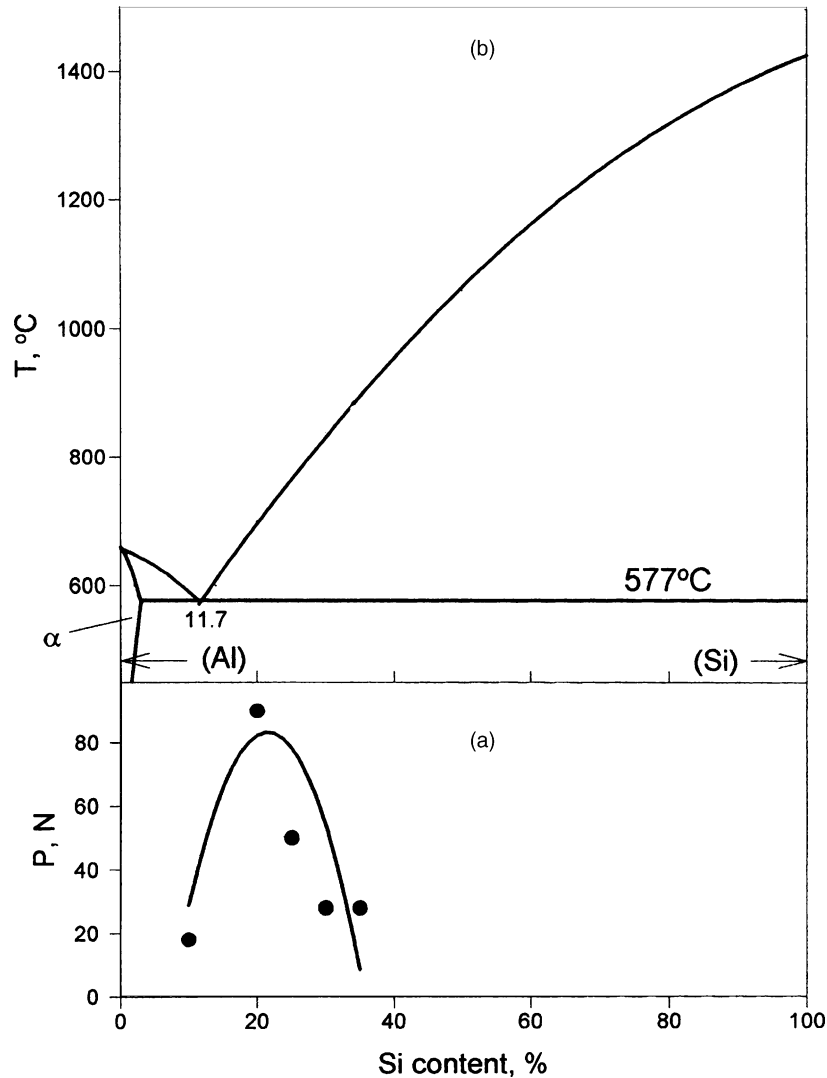


Fig. 8. (a) The yield point *P* as a function Si content. (b) The aluminum–silicon phase diagram (Si content in wt.%).

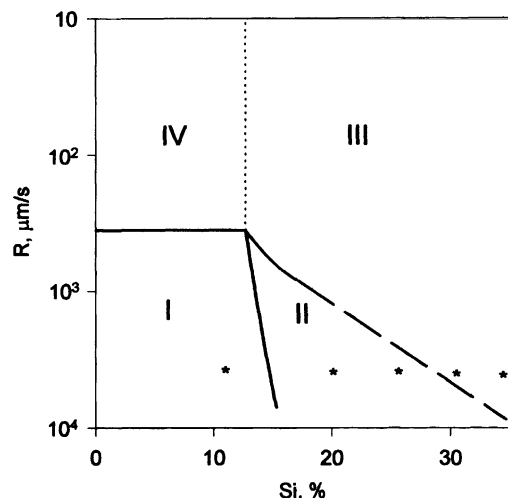


Fig. 9. Diagram of Al–Si alloys microstructure vs. Si concentration and cooling rate R , constructed from the data of [6], including extrapolation of the upper border for coupled growth from 17 to 35 wt.% Si. Region I corresponds to a fiber-like structure with primary crystals of α Si solid solution in Al. Region II corresponds to a fine-grained eutectic-like structure. Region III corresponds to a flake structure with primary Si crystals. The dotted boundary separates the regions where the primary crystals are alpha phase (region I) and silicon (region II). The star-like points indicate conditions used in the present work.

melt quenching was performed for similar conditions and the measured cooling rate supported our estimated value. The alloys compositions and the approximate solidification rate used in the present work are indicated by star-like points in Fig. 9. From this figure, it follows that our Al–20 wt.% Si alloy was produced in the region coupled eutectic growth and its finest eutectic-like structure without any primary crystals is responsible for the maximum yield point. The samples of Al–11.5 wt.% Si were obtained outside this region and had a coarser eutectic structure with α dendrites, resulting in lower yield point. The samples of 25 wt.% Si were cast close to the border of the region III and the alloys with 30 and 35 wt.% Si in the region III where primary Si crystals were growing. It is necessary to note that the borders of the regions in [11] were shown for 50% transformation of the structures. Therefore the alloys with Si content in the region from 25 to 35 wt.% for our the solidification rate should have had inclusions of primary Si crystals, the sizes and numbers of which increased with increasing Si content. The interface between these crystals and the eutectic matrix decreases the yield point.

Increasing the Si content is thought to decrease the temperature coefficient of Young's modulus due to increasing covalent contribution to atomic bonding. The average temperature coefficient of Young's modulus of Al from 20 to 500 °C is about $6.5310^{-4} \text{ K}^{-1}$ [14], while the temperature coefficient of Young's modulus of Si for the (1 0 0) direction is $7.7910^{-5} \text{ K}^{-1}$ [15]. Therefore, it is possible to assume that the maximum of $\Delta E/E\Delta T$ at 20 wt.% Si is caused by changing the microstructure of the alloy due to annealing during the measurement of temperature dependence of Young's modulus. In the quenched samples the Si content of α solid solution

exceeds the equilibrium value. The Guinier–Preston zones can arise under heating of the samples during these measurements. In the zones with higher Si concentration the contribution of covalent interatomic forces increases. It causes the increasing of E during the heating of samples and decreasing of the temperature coefficient of E . Maximum on the concentration dependence of the temperature coefficient of Young's modulus at 20 °C can be explained by the finest eutectic structure of the alloy making difficult the creation of the zones. The same process of the zones creation can explain also the temperature hysteresis of Young's modulus and minimum on concentration dependence of $\Delta E_{20}/E_{20}$.

It is known [11] that the addition of Sr to Al–Si alloys leads to finer-grained structures for all Si contents. This explains why Al–11.5 wt.% Si–0.01 wt.% Sr showed no hysteresis.

5. Conclusion

Al–Si alloys prepared by rapid cooling of levitated melts were used to elucidate the effect of Si content on mechanical properties. The structure and mechanical properties of quenched Al–Si alloys in a wide compositional range from 11.5 up to 35 wt.% Si were investigated. Extrema versus silicon concentration were found for the average temperature coefficient of Young's modulus for cooling from 500 to 20 °C, the hysteresis of Young's modulus for heating and cooling over this range, and the yield point. These extrema correspond to solidification in the region yielding coupled eutectic-like growth. In this region the process of Guinier–Preston zones creation is hampered. It explains extrema of concentration dependences of temperature coefficient and temperature hysteresis of Young's modulus. The decrease of yield point for lower Si concentrations is attributed not only to the decreasing size and concentration of Si in the eutectic structure, but also to an increased number of plastic dendrites of α solid solution. The decrease of the yield point of hypereutectic alloys with increasing of the Si content is attributed to the increase of the interfacial area between primary Si crystals and eutectic matrix.

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References

- [1] G.B. Stroganov, V.A. Rotenberg, G.B. Gershman, Aluminum–Silicon Alloys, Metallurgiya, Moscow, 1977 (in Russian).
- [2] M. Gupta, S. Ling, J. Alloys Compd. 287 (1999) 284–294.
- [3] S.P. Nikanorov, V.V. Peller, in: G.E. Totten, D.S. Mackenzie (Eds.), Handbook of Aluminum, Physical Metallurgy and Processes, vol. 1, Marcel Decker, New York, 2003, pp. 81–211.

- [4] L.F. Mondolfo, *Aluminum Alloys: Structure and Properties*, Butterworths, London, 1976.
- [5] P. Li, V.I. Nikitin, E.G. Kandalova, K.V. Nikitin, *Mater. Sci. Eng. A* 332 (2002) 371–374.
- [6] K. Kocatepe, C.F. Burdett, *J. Mater. Sci.* 35 (2000) 3327–3335.
- [7] R. Alireza, M. Kenji, Toshiyuki, *Metall. Mater. Trans. A* 29 (1998) 1477–1484.
- [8] R. Alireza, K. Miwa, *Metall. Mater. Trans. A* 31 (2000) 755–762.
- [9] R.P. Liu, D.M. Herlach, M. Vandyoussefi, A.L. Greer, *Metall. Mater. Trans. A* 35 (2004) 607–612.
- [10] Yu.A. Burenkov, S.P. Nikanorov, *Phys. Solid State* 44 (2002) 318–323 (transl.: *Fizika Tverdogo Tela* 44 (2002) 307–311).
- [11] D.C. Jenkinson, L.M. Hogan, *J. Cryst. Growth* 28 (1975) 171–187.
- [12] O.P. Bobrov, S.N. Laptev, V.A. Khonik, *Fizika Tverdogo Tela* 46 (2004) 457–461 (*Solid State Phys.* 46 (2004)).
- [13] J. Mahmoud, H. Frederiksson, *J. Mater. Sci.* 35 (2000) 4947–4987.
- [14] G. Simmons, *Single Crystal Elastic Constants and Calculated Aggregate Properties*, J. Graduate Research Center, Southern Methodist University Press, Dallas, TX, USA, 1965.
- [15] Yu.A. Burenkov, S.P. Nikanorov, *Solid State Phys.* 16 (1974) 963–965 (transl.: *Fizika Tverdogo Tela* 16 (1974) 1496–1498).