


NEW METHODS OF OBTAINING Al-Si ALLOYS USING AMORPHOUS MICROSILICA

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Abstract

The work presents the calculations of enthalpy of formation as well as Gibbs energy of reduction of silicon dioxide by aluminum. This way, the possibility of producing Al-Si alloys using amorphous silica fume has been established. The influence of alloying additives and impurities on the silicon reduction process has been determined. The possibility of using magnesium as a surface-active additive, which allows for the removal of oxygen from the surface of dispersed particles and the reduction of silicon, has been demonstrated. Effective methods for producing hypoeutectic and hypereutectic Al-Si alloys using amorphous silica fume have been developed. Alloys with a silicon content exceeding 6 wt% were obtained by blowing preheated ($t = 300\text{ }^{\circ}\text{C}$) amorphous silica fume into the aluminum melt ($t = 900\text{ }^{\circ}\text{C}$) along with the stream of argon followed by intense mixing. Alloys with a silicon content exceeding 16 wt% were manufactured by induction melting of a silicon-containing mixture (60% SiO_2 , 40% Al + 20%

$3\text{NaF}\cdot 2\text{AlF}_3$) subjected to the presintering ($t = 800\text{ }^{\circ}\text{C}$) when the amorphous silica fume was reduced to crystalline silicon. It has been established that crystalline silicon, which is formed during the roasting of the tableted burden (as a result of reactions in solid phases), is smoothly absorbed by the aluminum melt. Aluminum oxide, obtained during the redox reaction, dissolves in cryolite, after which aluminum and silicon are fused together and transferred to the melt. Industrial implementation of the proposed methods will result in the improved efficiency of the existing process of manufacturing Al-Si alloys, as the energy-intensive stage of the crystalline silicon production is partially excluded, as well as reduce adverse environmental impact.

Keywords: aluminum, silicon, Al-Si alloys, silumins, microsilica, silicon dioxide

Introduction

Alloys of Al-Si system are the most common aluminum casting alloys. They represent a group of aluminum alloys with the main alloying element being silicon. The demand for Al-Si alloys is provided by a unique combination of their basic properties: low density, high fluidity, relatively low shrinkage, low tendency to formation of stresses and

cracks, high values of strength properties, wear resistance and heat resistance.^{1,2} Therefore, these alloys are widely used in automotive and aerospace industries, sometimes in critical applications.

Foundry Al-Si alloys can be produced by electrothermic, metallothermic, and electrolytic methods, as well as by dissolution of crystalline silicon in the aluminum melt.³⁻⁵

The first three methods are one-stage (i.e., alloys are produced of silica as a raw material in ore-thermal furnaces or reduction cells); however, due to the high power consumption for their implementation, difficulty of producing alloys of a given composition, as well as a possibility of alloy contamination by impurities and nonmetallic inclusions, these methods have not found wide industrial application.⁶

Today, the main Al–Si alloys ingot production method is crystalline silicon dissolution in the aluminum melt in aluminum smelters. The main advantage of this method is its high performance and the ability to produce alloys with the specified silicon content. However, this method has some significant disadvantages: high metal waste due to burning loss, low absorption of silicon fines (less than 5–6 mm) and high power consumption.

In the conditions when aluminum and silicon industries are located in close proximity, a method of producing Al–Si alloys using 30–50% Al–Si liquid alloy can also be employed. The method consists of pouring molten silicon into a vacuum ladle which contains aluminum melt. Despite high-quality alloys produced, the method is usually associated with logistical difficulties.

Thus, production of alloys of the Al–Si system under the existing layout assumes the presence of two metallurgical productions: primary aluminum and crystalline silicon, which entails high economic costs and power consumption. At the same time, silicon production generates wastes, namely dust of the gas treatment systems of the electrothermic furnaces, consisting of 85–95 wt% of micro- and nanoparticles of amorphous silicon dioxide. The dust yield of the silicon production ranges from 300 to 1000 kg per 1 t of commodity silicon.

Currently, most metallurgical production facilities accumulate the obtained microsilica in the slurry fields. This raises substantial economic losses, firstly, due to the costs of waste storage and burial, and, secondly, with lost profits from their industrial applications.

Use of silicon dust as a source of silicon may improve efficiency of the Al–Si alloys production process by partial exclusion of the energy-intensive metallurgical silicon production stage.

The price of crystalline silicon, depending on the brand, varies from \$2000 to \$2500 per ton. Dust emitted from the gas systems involved in the cleaning of electrothermal furnaces can be used in the production of Al–Si alloys, as it is associated with a very low cost (~ \$25 per ton) and increased energy efficiency which leads to a significant economic effect can be achieved. Moreover, recycling and use of silicon dust wastes should be seen as an important

way of saving material resources and increasing environmental safety of the adjacent territories.^{1,7}

In excess of described methods and patents designed to obtain Al–Si alloys, there seems to exist no technologies that can be practically applied in manufacturing sector.^{1–9} Hence, the purpose of the work is to research a complex of physical–chemical processes of Al(l)–SiO₂ interface and develop a technology for producing low-temperature Al–Si alloys based on amorphous microsilica obtained from the silicon waste production.

Research Methods

The calculation of the thermodynamic probability of silicon reduction by aluminum from amorphous microsilica, as well as calculations of the stability of chemical compounds formed during the interaction of silicon oxide with technical aluminum impurities, was carried out over a wide temperature range—based on data on the basic physico-chemical parameters (standard enthalpy of formation, Gibbs free energy), as well as using the programs “PANDAT,” “CHS Chemistry 5” and “SELECTOR.”⁹ Since the initial data on thermodynamic functions standard values for a number of the considered chemical compounds do not exist (in the literature and used software) or not correlates with each other, well-known and adapted methods of approximate calculations were used.^{10,11}

Commercial purity aluminum with the following content of the main impurities (wt%: Si—0.10, Fe—0.112, Mg—0.02, Mn—0.023, Cu—0.015, Zn—0.031, Ga—0.001, Ti—0.001, V—0.001) was used as the base metal for the laboratory research on Al–Si alloys production with application of amorphous silica.

Flotation treatment for silicon production dust makes it possible to separate silicon dioxide particles of a size up to 100 nm (Figure 2).^{12,13}

The introduction of silica into molten aluminum was accomplished by different methods:

- (1) to the melt surface;
- (2) to the bottom of the crucible followed by melt pouring;
- (3) using a mechanical bolder (in a bell);
- (4) by portions into a funnel formed with melt continuous stirring;
- (5) in the form of master alloy as pressed tablets of the composition aluminum powder—SiO₂;
- (6) placing particles in melt in a semiliquid condition.
- (7) by blowing SiO₂ particles as well as the stream of argon into the melt;

- (8) induction melting of the previously prepared mixture (60% SiO₂–40% Al–20% 3NaF·2AlF₃) under a layer of cryolite.

There are several factors that complicate Al–Si alloys production with the use of amorphous microsilica by its simple introduction in the aluminum melt:

- (1) low contact area of the microsilica powder with aluminum melt;
- (2) presence of gas films on the surface of the microsilica particles, preventing effective interaction at the interface of the liquid and solid phases;
- (3) presence of 40–50 vol% of air in the powder, reducing its density, heat capacity and thermal conductivity;
- (4) presence of microsilica particles in the developed surface and high surface energy.

In this regard, four methods were characterized by low efficiency, since their implementation failed to maintain the constant contact in the Al(l)–SiO₂ system as well as the transformation of silicon to aluminum melt.

The use of preformed ligatures ensures reduction of silicon; however, the process occurs directly in the contact zone of the melt with the surface of the tablet as well as in the tablet volume (upon the interaction of aluminum powder particles and silica fume).

The introduction of silica fume into the solid–liquid melt allows to obtain alloys with a silicon content exceeding 3 wt%, as the aluminum melt in its solid–liquid state is characterized by high viscosity which, in turn, facilitates the mixing in of dispersed particles.¹³

The greatest efficiency was demonstrated by the methods of injecting silicon dioxide into an aluminum melt in the

argon flow, as well as induction melting of a previously prepared silica-containing mixture. Silica injection into an aluminum melt was carried out in the laboratory setup shown in Figure 1.

Aluminum with the weight of 1000 g was placed in the crucible made of boronsiliconized graphite. The heat took place in top-charging muffle shaft furnace. The metal was heated to a temperature of 900 °C, melted, and then amorphous microsilica (previously subjected to heat treatment inside a heated bin) was introduced in it together with the inert gas stream. Figure 2 presents the SEM images of microsilica particles of a spherical shape. It is seen that the size of SiO₂ particles varies within a wide range, and smaller particles stick to the surface of the larger ones (due to the high surface energy of the latter) (Figure 2b).

To improve wettability of the microsilica particles and prevent formation of agglomerates, before introduction into the aluminum melt they were subjected to the heat treatment at a temperature of 200–300 °C. Likewise to solve the wettability problem the melt was alloyed with magnesium. Magnesium was introduced into the melt in the form of MG-90 magnesium alloy. The use of magnesium in the Al–Si alloys production process was conditioned by the fact that it can act as a surface-active additive in the melt for removal of oxygen from the surface of dispersed particles, decline of the surface tension of the molten aluminum and reduction in the energy of interfacial interaction between the solid and liquid phases.^{14,15}

To increase the degree of transition of silicon into the aluminum melt, the technology of producing Al–Si alloys using a previously prepared mixture consisting of silicon dioxide, aluminum powder and cryolite was tested. The purpose of the preliminary preparation of the mixture was the reduction of amorphous microsilica to crystalline

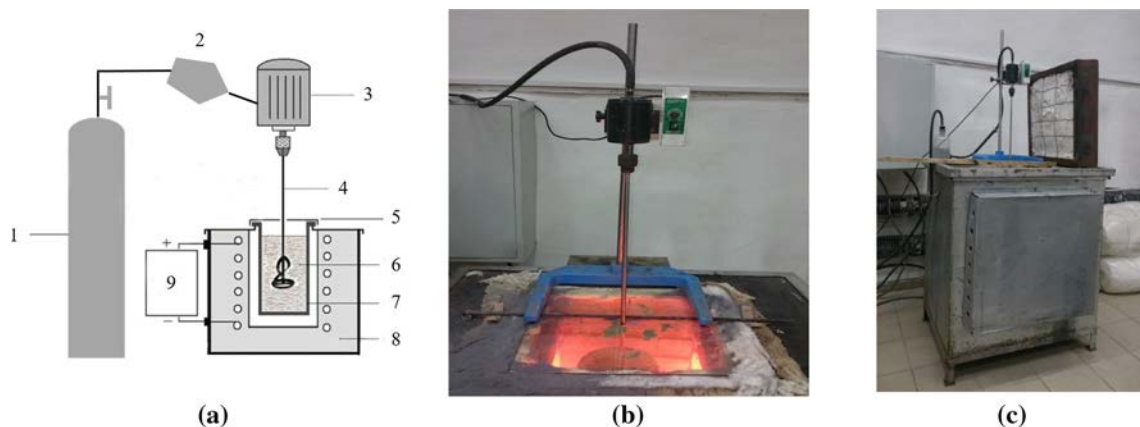


Figure 1. Scheme (a) and images (b, c) of the laboratory setup for producing Al–Si alloys: 1—vessel filled with argon; 2—heated bunker filled with silica fume; 3—laboratory stirrer; 4—hollow steel rod with impeller; 5—tempered glass (10 mm); 6—aluminum melt; 7—crucible (borosilicated graphite); 8—induction melting furnace IPP–5C; 9—thermostat.

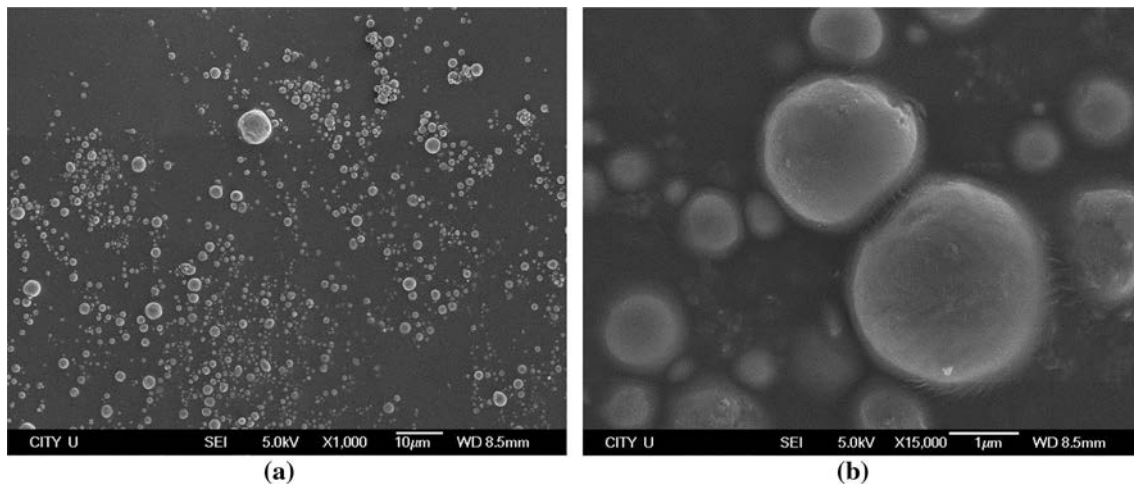


Figure 2. SEM images of silicon dioxide particles.

silicon. The procedure was performed out sequentially in several stages:

- (1) preparation of the mixture consisting of silicon dioxide (60 wt%), aluminum powder (or aluminum chips) (40 wt%) and low-modular cryolite (20 wt%) by weight of the main components. The ratio in the mixture of silicon dioxide and powdered aluminum is defined in accordance with the stoichiometric calculation 3SiO_2 (180 g/mol) + 4Al (108 g/mol) \rightarrow $2\text{Al}_2\text{O}_3$ (204 g/mol) + 3Si (84 g/mol). Cryolite is essential to the mixture as it creates an atmosphere of reducing roasting and shifting the reaction equilibrium toward the formation of the end products.
- (2) pressing the mixture into a tablet (with a force of 100 kN);
- (3) tablet roasting at the temperature of 800 °C for 30 min.
- (4) grinding the sintered material (using planetary ball mill RM 100) to a fraction of 100 μm, which ensures the maximum contact degree in the Al–Si system.

Upon the completion of the preparatory operations, the mixture containing silica in crystalline form was loaded into a crucible of borosilicate graphite.

Solid aluminum was placed in the crucible, filled with a five-centimeter layer of cryolite, corrected with aluminum fluoride to reach the melting point ($t = 800$ °C). The melting was carried out in the induction furnace at the temperature range of 850–900 °C. After the metal was melted, it was intensively mixed mechanically for 5 min.

To remove the by-products of the reduction reaction between silicon and aluminum, as well as nonmetallic

inclusions, gases and oxide films, metal was treated with the covering-refining flux before metal casting.

SiO_2 particles were mixed in the melt using IKA EURO-STAR 200 Control P4 laboratory mixer (Germany) with a rotation speed range of 0–530 rpm.

The silicon content in the metal before and after the experiments was determined using optical emission spectrometer with spectrum spark excitation source SPECTROLAB manufactured by SPECTRO Analytical Instruments (Germany).

To study the phase composition of the impurities included in the composition of the tested samples, the X-ray diffraction analysis was used with application of X-ray diffractometer XRD–7000 manufactured by Shimadzu. The metal samples were studied in the 2θ -range from 10° to 70°.

The microstructure of the samples was studied using scanning electron microscope JIB-4500 Multibeam manufactured by JEOL, equipped with energy-dispersive detector X-Max manufactured by Oxford Instruments. The microstructure of the samples was studied in the mode of secondary and back-scattered electrons.

For the thorough analysis of the microstructure of the samples of the obtained Al–Si alloys, metallographic examinations were carried out using an Olympus GX–51 inverted optical microscope (Japan). The macrostructure of the samples was studied using stereoscopic microscope SZX 16 manufactured by Olympus (Japan). Quantitative analysis of the parameters of macrostructure was performed using an image analyzer SIAMS 700 by SIAMS Ltd. (Russian Federation).

Results and Discussion

Study of the possibility for Al–Si alloys production by introduction of SiO₂ particles in the aluminum melt is impossible without a preliminary thermodynamic analysis. The work determined the thermodynamic possibility for proceeding of the process of silicon reduction by aluminum from amorphous silica ($4\text{Al} + 3\text{SiO}_2 \rightarrow 2\text{Al}_2\text{O}_3 + 3\text{Si}$) in the temperature range between 298 and 1600 K (25–1327 °C). The values of Gibbs free energy in the investigated temperature range were calculated with a step of 100 K. The calculations were carried out upon considering polymorphic transformations of silica (quartz–tridymite–cristobalite) in accordance with the existing data.^{16,17}

It was found that Gibbs free energy has negative values throughout the investigated temperature range (Figure 3).

Relationship $\Delta G - T$ is shifted to the region of less negative values with temperature rise. The relationship between Gibbs free energy and temperature is represented by two straight lines crossing at the point corresponding to the aluminum melting temperature (933 K or 660 °C). Angles of inclination of these straight lines differ from each other insignificantly. This shows, first, a possibility of amorphous silica reduction at the temperatures above the aluminum melting temperature, and, secondly, the reduced intensity of this process proceeding at subsequent higher temperatures.

It was also discovered that polymorphic transformations of silica, which occur at the temperatures of 1168 and 1543 K (895 and 1270 °C), do not affect the smoothness of the $\Delta G - T$ curve.

To determine the effect of alloying additives and impurities contained in technical aluminum and silicon dust on the process of silicon reduction, the analysis of physicochemical parameters characterizing the thermodynamic stability of the compounds formed in a wide temperature range was

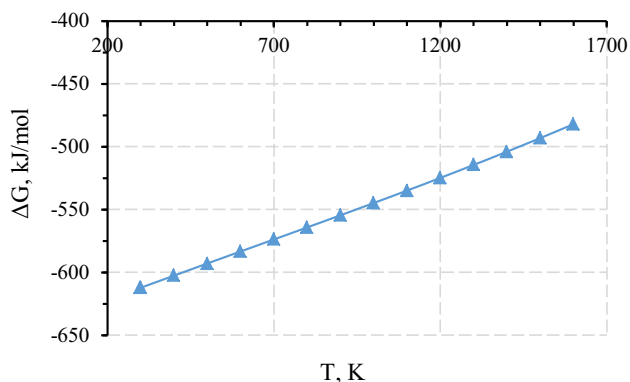


Figure 3. Change of Gibbs energy of the reaction of silicon oxide interaction with aluminum, depending on the temperature.

employed as the main research method. Among the chemical compounds formed in the Al–SiO₂ system, Al₂O₃, Al₄C₃, SiC, Mg₂Si and MgO proved to be of the utmost interest with regard to their impact on the interfacial interaction process in the system.

The calculations were carried out within a wide range of temperatures. The standard temperature of 298 K was picked as the initial one while the maximum melting point of the chemical compound (~ 2000–3000 K or 1798–2798 °C) was to be the final one. Al₄C₃ and SiC were chosen as carbon is present as a constant aluminum impurity, and silicon production dust contains up to 25 wt% of C. The formation of magnesium compounds (Mg₂Si, MgO) was studied as magnesium was used as a surfactant to increase wettability in the Al(l)–SiO₂ system during the experimental research

According to the calculations, the Gibbs free energy in all compounds has negative values in the studied temperature ranges under study (Figure 4).

This phenomenon suggests that during the production of silumins using amorphous silica fume as a source of silicon in the temperature range of 690–900 °C, additional refining of the aluminum melt with the use of fluxes is deemed necessary. The stability of the MgO chemical compound indicates the possibility of using magnesium as a surface-active additive that helps to remove oxygen from the surface of the dispersed particles and reduce silicon from silicon dioxide, along with aluminum.

Production of Al–Si Alloys by Blowing Amorphous Silica into an Aluminum Melt in the Argon Stream

The alloy samples obtained in the course of experimental studies were examined by means of optical emission spectroscopy, scanning electron spectroscopy, optical microscopy, as well as by X-ray diffraction.

The results of optical emission spectroscopy demonstrate that in the resulting alloy compared to the original aluminum, the silicon content increased from 0.112 to 6.358 wt% (Table 1).

The diffraction pattern of the sample belonging to the produced alloy exhibits the peaks corresponding to aluminum, silicon and SiO₂ compounds (Figure 5).

The peaks corresponding to metallic aluminum ($2\theta = 44.7^\circ$; 65.2°) and crystalline silicon ($2\theta = 26.3^\circ$) possess the highest intensity. The peaks belonging to silicon dioxide are present in the 2θ -range from 20° to 30° . Their weak intensity is due to predominance of silicon in the crystalline form in the structure of the produced alloy.

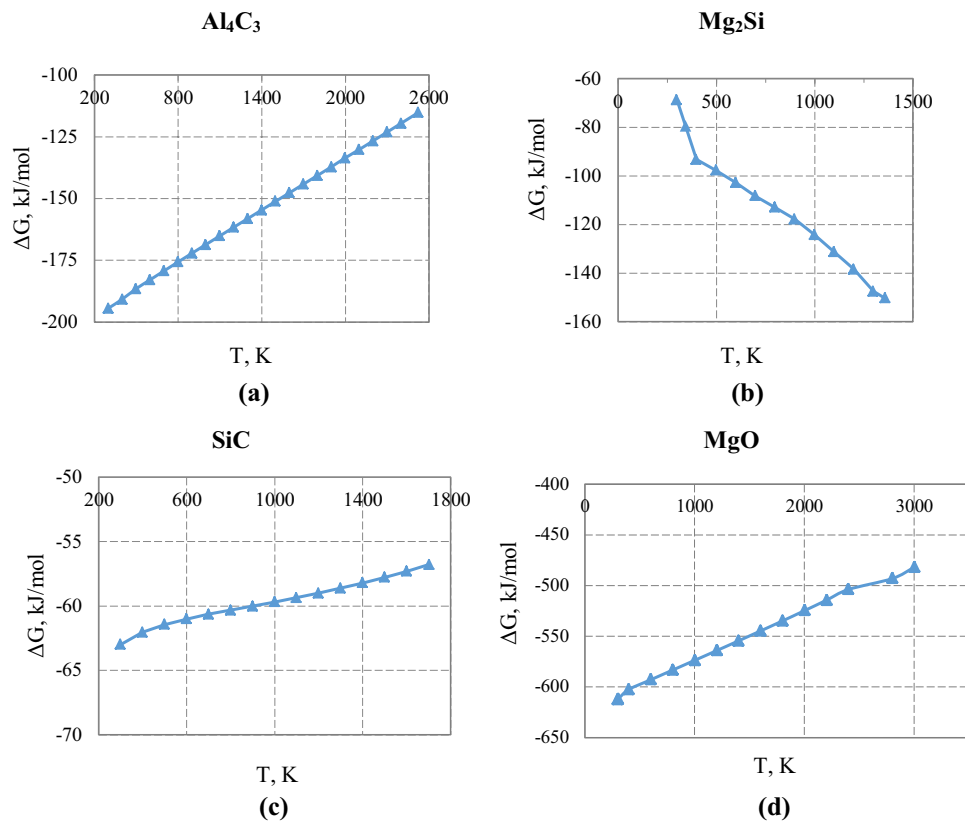


Figure 4. Change of Gibbs energy of the following compounds: (a) Al_4C_3 ; (b) Mg_2Si ; (c) SiC ; (d) MgO .

Table 1. The Content of Impurities in the Original Aluminum and the Resulting Alloy

Content of impurities, wt%	Si	Fe	Mg	Mn	Zn	Cu	Ga	Ti	V
The original aluminum	0.110	0.112	0.020	0.023	0.031	0.015	0.001	0.001	0.001
The resulting alloy	6.358	0.338	0.020	0.027	0.031	0.015	0.001	0.001	0.001

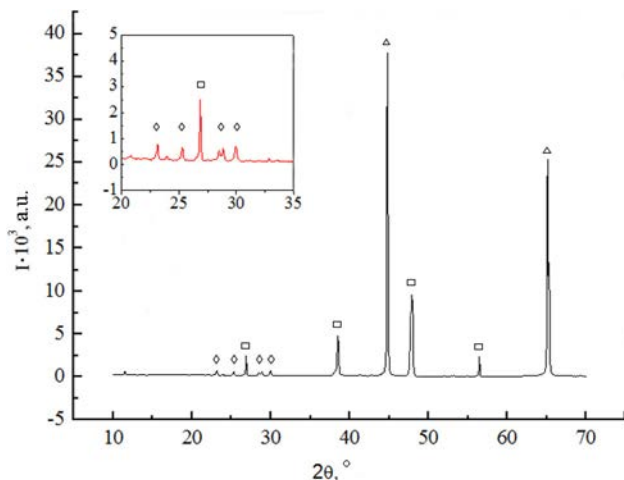


Figure 5. The diffraction pattern of the produced alloy in the 2θ range of $10\text{--}70^\circ$ (triangle—peaks of Al; square—peaks of Si; diamond—peaks of SiO_2).

The studies of the microstructure of the samples of the original aluminum and the resulting alloy confirmed the increase in the silicon content in the resulting alloy.

The energy-dispersive detector of the microscope during scanning of the electron beam made it possible to record the intensity of the characteristic X-ray lines of the elements and, consequently, to receive their two-dimensional map of distribution. The distribution maps for the basic elements (Al, Si) are presented in Figure 6.

The distribution map for silicon of the produced alloy sample clearly shows the presence of silicon inclusions at the aluminum grain boundaries (Figure 6d), which are virtually absent in the original metal (Figure 6b). This confirms the results obtained during spectral analysis of the samples.

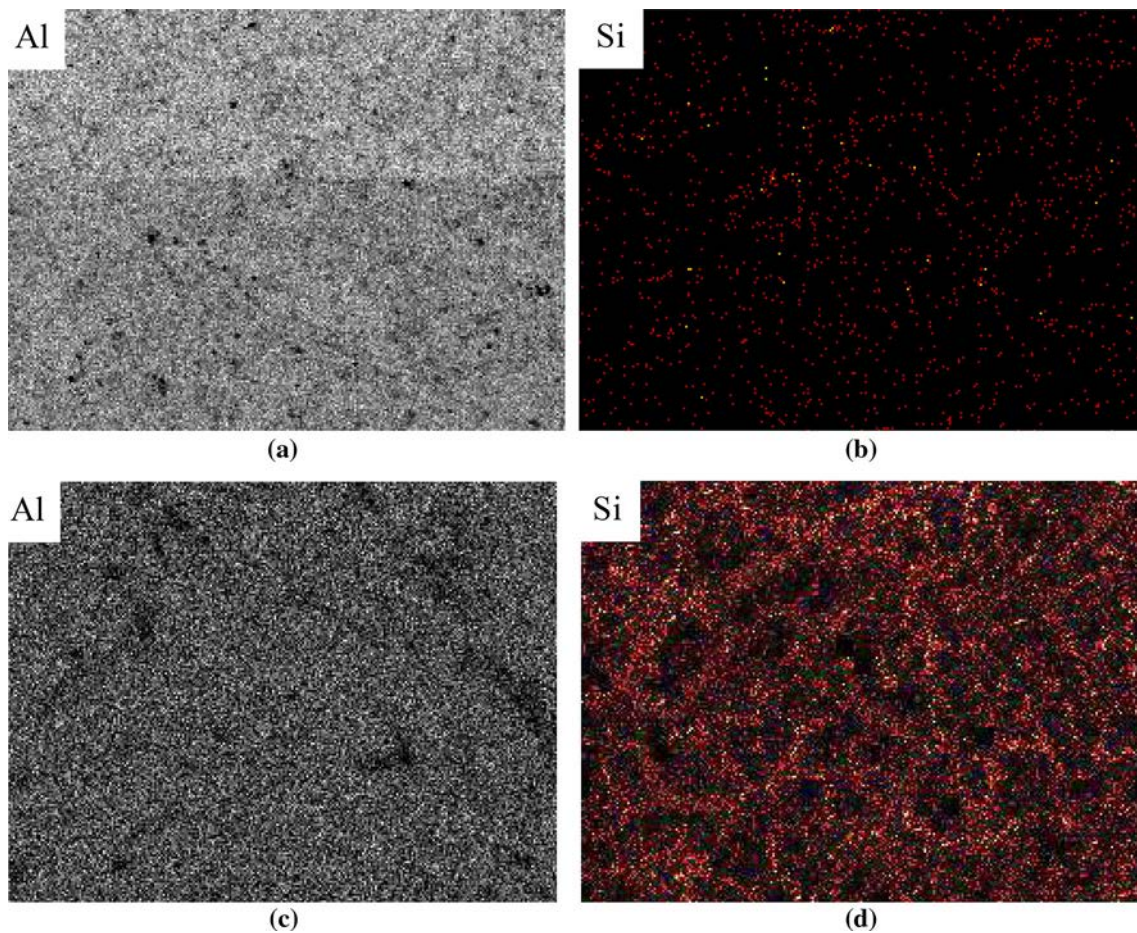


Figure 6. The distribution maps for aluminum (a, c) and silicon (b, d) of the samples of the original aluminum and the resulting alloy (the bright regions on the images correspond to Al and Si atoms).

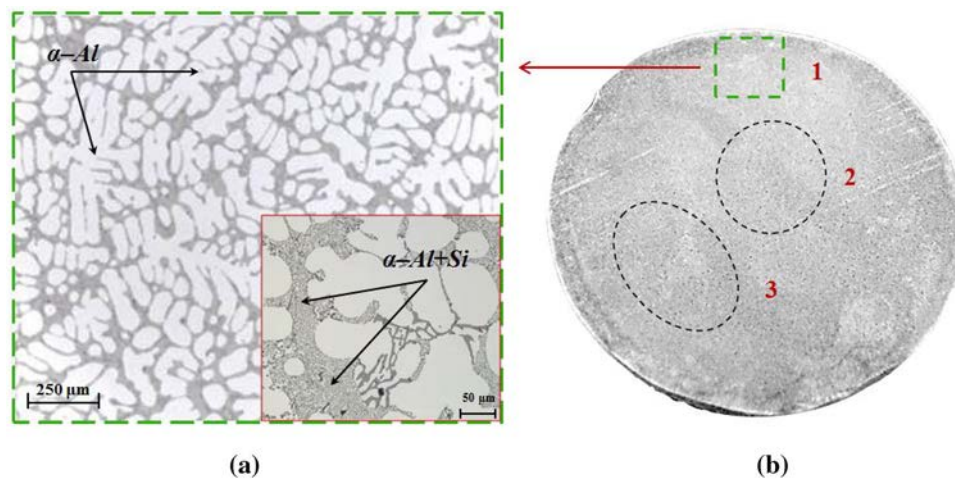


Figure 7. Micro- (a) and macrostructure (b) of the resulting alloy (1—area with representative microstructure; 2,3—areas with scattered porosity).

The produced alloy microstructure corresponds to the hypoeutectic Al–Si alloys and consists of dendrites of a solid solution of silicon in aluminum ($\alpha - \text{Al}$) and eutectic $\alpha - \text{Al} + \text{Si}$, located in the interdendritic space (Figure 7a). In course of the analysis of alloy microstructure, it

was not detected any no iron-containing phases. It is possible to observe a cross section (macrograph) of the produced ingot in Figure 7b. It should be noted that the ingot micro- and macrostructures are characterized by availability of a scattered porosity, the occurrence of which is

due to the presence of SiO₂ particles, which have not started interacting with aluminum, in the volume of the crystallized metal.

Despite the development of a scattered porosity in the casting (density index—4.9%), the conducted research showed the possibility and the prospects of Al–Si alloys production with use of amorphous silica as the silicon source. A preliminary thermal treatment of amorphous silica had a decisive influence on the intensity of interaction in Al(l)–SiO₂ system; it allowed removal of gas films from the surface of its particles, as a result, to reduce the surface tension of the melt, as well as to reduce the energy of interfacial interaction between the solid and liquid phases. Acceleration of the process of silicon reduction from its oxide was provided by introduction of silica particles together with a stream of argon and their subsequent vigorous stirring.

Production of Al–Si Alloys by Induction Melting of a Previously Prepared Silica-Containing Mixture

The studies demonstrated that upon injecting amorphous silica into aluminum melt in the argon stream, only hypoeutectic alloys can be obtained. Such a phenomenon can be explained by both physical properties of the silica (amorphous structure, particle size, density) and the technological design of the process (e.g., cooling of the melt temperature during its long purging with argon). Further, it was discovered that to obtain eutectic and hypereutectic Al–Si alloys, preliminary preparation of the silicon-containing mixture, including pressing it (60% SiO₂–40% Al–20% 3NaF·2AlF₃) into tablets, subjecting it to the heat treatment ($t = 800$ °C) and the subsequent grinding of the sintered mass to a fraction of 100 microns are required.

The choice of sintering temperature is determined by the need to create conditions for the solid-phase interaction. The grinding of the sintered mass is supposed to increase the area of its contact with the aluminum melt. As a result of sintering of the silica-containing mixture, practically all of its microsilica (more than 95 wt%) are reduced to crystalline silicon.

The bulk weight of the resulting mixture was 1.4–1.6 g/cm³. This value does not exceed the density of the molten low-modulus cryolite, which makes it possible to create a

mechanical mixture of these components. In the cryolite environment, aluminum oxide dissolves, and aluminum and silicon are melted. For the completion of the process, the optimum concentration of Al₂O₃ in cryolite amounting to 10–11 wt% was ensured.^{6,7}

Crystalline silicon formed during tablet roasting (as a result of reactions in solid phases) is smoothly absorbed by the aluminum melt. Aluminum oxide, obtained as a result of oxidation–reduction reaction, dissolves in cryolite. After that, aluminum and silicon are melted.

The dissolution of alumina in cryolite can be explained by the exchange of F⁻ and O²⁻ ions between the anions of AlF₆³⁻ of molten cryolite and the alumina lattice. The cryolite Al³⁺ cations characterized by the strong field pull the O²⁻ out of the alumina lattice. The exchange causes the breakage of the alumina crystal lattice which, in turn, leads to the dissolution of the alumina. Thus, aluminum cations Al³⁺—the components of the cryolitic complexes AlF₆³⁻ and AlF₄⁻, serve as “solvents” of alumina in cryolite.

In the course of experimental research, a crucible made of borosilicated graphite was loaded with 780 g of the mixture (28% crystalline silicon, 50% aluminum oxide and 20% cryolite), 1000 g of A7 grade aluminum and 500 g of cryolite. As a result of experimental melting, a hypereutectic alloy was obtained ($m = 1340$ g) where the silicon content exceeded 16 wt%. That was confirmed by the results of optical emission spectroscopy (Table 2).

The results demonstrate that practically all crystalline silicon and aluminum that were in the mixture in the form of Al₂O₃ turned into the melt. The use of a laboratory stirrer with a steel impeller can be accounted for the increase in the iron content in the resulting alloy. To reduce the iron content in the alloy and prevent the reduction in the complex of mechanical and casting properties, the steel impeller can be replaced with titanium one and mixing can be carried out by an electromagnetic method. Cryolite from smelting was discharged separately for further recycling.

Thus, the possibility of almost complete absorption of silicon from the composition of amorphous silicon dioxide in an aluminum melt and the production of hypoeutectic and hypereutectic Al–Si alloys were experimentally proved.

Table 2. The Content of Impurities in the Original Aluminum and the Resulting Alloy

Content of impurities, wt%	Si	Fe	Mg	Mn	Zn	Cu	Ga	Ti	V
The original aluminum	<i>0.110</i>	0.112	0.004	0.023	0.011	0.009	0.001	0.008	0.005
The resulting alloy	<i>16.40</i>	1.072	0.004	0.085	0.012	0.009	0.001	0.008	0.005

Italic values emphasize that the content of silicon greatly increased

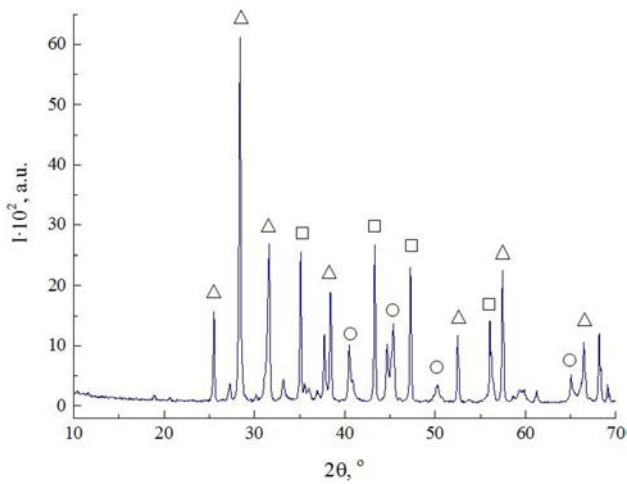


Figure 8. The diffraction pattern of the produced alloy in the 2θ range of $10\text{--}70^\circ$ (triangle—peaks of Al; square—peaks of Si; circle—peaks of FeSiAl_5).

The diffraction pattern of the sample belonging to the produced alloy exhibits the peaks corresponding to aluminum, silicon and intermetallic compounds based on iron (Figure 8).

The peaks corresponding to metallic aluminum ($2\theta = 26.3^\circ; 28.1^\circ; 32.4^\circ; 38.5^\circ; 57.9^\circ; 68.1^\circ$) and crystalline silicon ($2\theta = 35.2^\circ; 44.7^\circ; 47.3^\circ; 56.2^\circ$) possess the highest intensity.

Due to the high degree of doping of the alloy, peaks of the iron-containing phase $\beta\text{-FeSiAl}_5$ ($2\theta = 41.8^\circ; 45.9^\circ; 50.7^\circ; 65.8^\circ$) are shown in the diffractogram—all in accordance with the existing data.^{18–23}

Optical microscopy of the samples of the obtained alloy confirmed that the microstructure of the obtained alloy corresponds to hypereutectic Al–Si alloys (Figure 9).

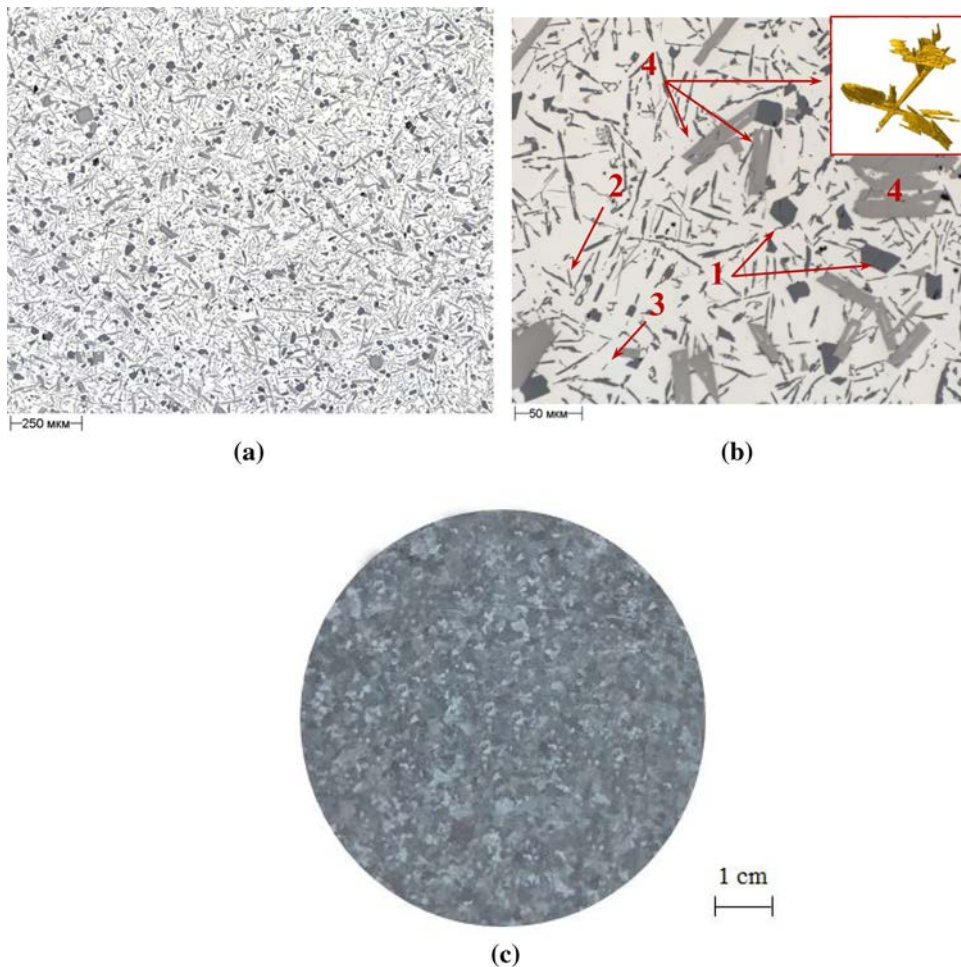


Figure 9. Micro- (a,b) and macrostructure (c) of the obtained hypereutectic Al–Si alloys (1—primary silicon crystals; 2—eutectic silicon crystals; 3—secondary silicon crystals; 4—iron-containing phases $\beta\text{-FeSiAl}_5$).

The macrostructure of the resulting alloy is homogeneous and demonstrates the absence of gas porosity in the mold (density index—2.3%). The sample structure is fine-grained—the average grain size is 0.85 mm.

Microstructure consists of the following components:

- crystals of the primary silicon in the form of faceted plates (average size of 10 μm);
- eutectic silicon crystals;
- fine equiaxed particles of the secondary silicon (size of up to 1 micron);
- large inclusions of iron-containing phases $\beta\text{-FeSiAl}_5$, which are complex 3D plate structures (insert in Figure 9b) and on the polished 2D specimen having the form of plates 1–8 μm thick and 10–50 μm long.

Second-order dendritic cells in the Al–Si alloy structure are not present due to the high iron content.

Conclusion

The calculations of Gibbs energy of the process of silicon reduction by aluminum from its oxide ($T = 298\text{--}1600\text{ K}$ or $25\text{--}1327\text{ }^\circ\text{C}$), made in the research, allowed establishing a possibility of producing alloys of the Al–Si system using microsilica amorphous particles as an alloying addition.

In the course of determining the influence of alloying additives and impurities on the reduction of silicon, it was demonstrated that magnesium can be used as a surfactant to remove oxygen from the surface of dispersed particles and to reduce silicon from the oxide.

The experimental studies have shown the prospects for production of Al–Si alloys by introducing the amorphous silica together with argon stream into the aluminum melt (which is at a temperature of $900\text{ }^\circ\text{C}$). Despite the low degree of wettability of SiO_2 particles, the necessary degree of their assimilation by the molten aluminum was achieved due to their pre-heat treatment.

Moreover, an effective method was proposed for obtaining branded hypereutectic alloys by the method of induction melting of a silica-containing mixture, during the preliminary sintering of which amorphous silica fume was reduced to crystalline silicon. The method allowed to obtain Al–Si alloys with a silicon content of 16.40 wt%. In this regard, it can be concluded that the presented methods for producing Al–Si alloys using amorphous silica fume as a source of silicon are effective, since they allow to involve silicon production in the process as well as produce cast alloys with minimal cost. Upon scaling up of technologies, the proposed methods can be implemented in the process of production of cast aluminum alloys. The use of silicon

production dust as a source of silicon will make it possible to increase the efficiency of the existing technological process of obtaining Al–Si alloys due to partial exclusion of the energy-intensive stage of production of metallurgical silicon.

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