MODERN CAST ALUMINUM ALLOYS

V. S. Zolotorevskii

UDC 669.715:669.018.28

Characteristic features of the structure and properties of cast aluminum alloys are discussed. Ways of elevating the ductility and fracture toughness of silumins, and methods of improving the casting properties of high-strength and corrosion-resistant silicon-free alloys, as well as possibilities of creating alloys with a high iron content from a mixed secondary raw material are demonstrated.

In the 1980s, the Soviet Union produced more than a million tons of shaped castings formed from aluminum alloys. There is no accurate statistical data for the past two years, but approximate assessment indicates that the overall output of aluminum and its alloys in the form of pigs and castings has diminished negligibly in Russia. The export portion of shaped materials, however, has increased significantly; as a result, the output of aluminum castings has most likely diminished significantly in Russia. Approximately half of the castings are produced from primary alloys, and the remainder from secondary ones. The overwhelming portion of the casting volume has gone into silumins (probably no less than 90%), while all other aluminum alloys are used, as before, on a limited scale.

The portion of secondary alloys produced from scrap and waste in the manufacture of shaped castings has increased steadily in recent years. This trend was intensified in 1991-92 as a result of a sharp increase in the export of pure aluminum (other than wastes). Here, the quality of secondary raw material that has gone into reprocessing has diminished: it has increasingly become a mixture of various alloys, including nonaluminum ones (primarily iron-base).

Casting formed from aluminum alloys are used chiefly as structural components. In this connection, the mechanical and anti-corrosion properties are the basic ones characterizing the service reliability and remaining life of the castings. No less important (especially for nonsilumins) are the casting characteristics, primarily the hot shortness and fluidity. As we know, silumins posses good casting, but not very high service properties. Among the nonsilumins, conversely, there are many alloys with very good mechanical and anti-corrosion characteristics, but they are all considerably inferior to silumins in terms of casting properties.

Table 1 indicates the composition and gives the properties of the cast aluminum alloys that are most widely used in industry and that have the best set of service characteristics.

As in the past, the familiar primary silumins AK12 (AL2), AK9ch (AL4), and AK7ch (AL9), their secondary analogues, and the AK5M2-type copper silumins remain the highest "multitonnage" alloys. Despite low mechanical characteristics, the high casting properties place them above the competition in many cases. Alloy AK8M3ch (VAL8) has the highest level of mechanical properties among the silumins. In silicon-free alloys based on the Al−Cu, Al−Mg−Zn, and Al−Zn−Mg−Cu systems, it is possible to achieve even higher values of strength, plasticity, ductility, high-temperature-strength, and other characteristics (Table 1, alloys AM4.5Kd, and VAL12). The casting properties of these alloys, however, are as before, considerably lower than those of the silumins. Similar results are also obtained for cast magnaliums with very high corrosion resistance (Table 1, alloy AMg10).

In our opinion, three basic problems should be solved for further development of cast aluminum alloys:

— improvement in plasticity and ductility (characteristics of rupture strength) in primarily silumins, especially copper silumins;

— improvement in the casting properties of nonsilumins without a reduction in the already attained high level of service characteristics, i.e., mechanical and anti-corrosion properties; and,

---

TABLE 1

Alloy*	Average content of basic alloying elements, %	Heat treat-ment	σ_u N/mm²	δ, %	HB	K_{Ic} N/mm$^{3/2}$	" ·I· N/mm²	H	F
								mm	
AK12 (AL2)	11,5 Si	—	150	4	50	—	40	<2,5	420
AK9ch (AL4)	9 Si; 0,2 Mg	T6	240	3,5	70	—	75	<2,5	360
AK7ch (AL9)	7 Si; 0,3 Mg	T6	240	1	70	—	45	5	350
AK5M2	5 Si; 2,5 Cu; 0,5 Mg	T5	210	0,5	75	18	55	12,5	340
Amg10	10 Mg;each),1 Ti, Zr, Be	T5	320	12	75	—	60	17,5	270
AK8M3ch (VAL8)	8 Si; 3 Cu; 0,15 Ti; 0,7 Zn; 0,1 Be; 0,05 B	T5	400	4	110	22	110	<2,5	360
AM4.5Kd (VAL10)	5 Cu; 0,6 Mn; 0,2 Ti; 0,15 Cd	T6	500	4	120	26	120	27,5	—
VAL12**	5,7 Zn;2,5 Mg; 1,5 Cu; 0,2 Mn; each,15 Ti; Zr; 0,1 Be; 0,15 Cr; 0,01 B	T6	550	3	130	25	90	22,5	—

*According to GOST 1583-89.

**On standard alloys.

Note. The hot-shortness indicator H was determined from a circular semicocked sample meeting the requirements of the All-Union Scientific-Research Institute of Aviation Materials, and the fluidity F from a rod sample.

Notation: T5 indicates quenching and incomplete artificial aging; T6 implies quenching and complete (for maximum strength) artificial aging.
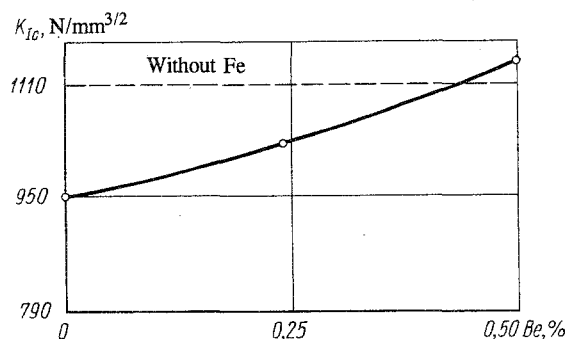

Fig. 1. Curve of fracture toughness K_{Ic} of Al + 7% Mg + 3% Zn + 1% Fe-alloy versus beryllium content.

— the creation of secondary alloys with satisfactory properties, which could be smelted from a low-grade mixed raw material.

Let us examine ways of solving these problems.

To provide good technological properties, cast alloys should contain a sufficiently large amount of eutectic components. In that case, the alloys have a narrow effective crystallization interval. Broad molten intergrain interlayers are observed in their structure during crystallization. This promotes the formation of minor shrinkage in the castings during hardening. Here, the deformation capacity of the alloys is high in the solid–liquid state. These alloys are not very prone to hot-crack formation [1]. The narrow crystallization interval predetermines the directivity of crystallization, and, consequently, the good fluidity of the alloys.

It should be pointed out that a high eutectic content in aluminum alloys predetermines the existence in them of a large amount of brittle excess phases, which are retained in the structure as relatively coarse particles even after heat treatment. The reduced plasticity and crack resistance of aluminum alloys with high casting properties (primarily silumins) is associated precisely with the brittle crystals of these phases. The brittle silicon phase is already contained in binary silumins, and the volume fraction of the brittle components becomes even greater when additives or impurities of copper, magnesium, iron, and other elements are present in commercial silumins. Embrittlement as a result of decomposition (primarily in terms of magnesium and copper) is, moreover, observed in multicomponent silumins. A reduction in the plasticity of castings also gives rise to inevitable pores of shrinkage and gaseous origin, and also to nonmetallic inclusions, primarily oxides. As a result, the relative elongation of standard copper-free silumins and copper silumins in tension is δ ≤ 4% and δ ≤ 1%, respectively. In that case, the fracture toughness K_{Ic} ≤ 6340-7930 N/mm$^{3/2}$ (Table 1).

TABLE 2

Alloy	Element content,* %					σ_u	$\sigma_{0,2}$	δ, %	K_{Ic},
	Si	Fe	Cu	Mg	Mn	Nm/mm²			N/mm³/²
						Cast state			
AK2Zh2	1,5—2,5	1,5—2,5	0,5—1,2	0,5—0,9	0,5	200—220	100—120	3—8	600—760
Ak2Zh2M2	2,5—3,5	2—2,8	2—3,5	0,2—0,6	0,2—0,8	210—230	120—140	3—5	505—665
AK2Zh2M7	1,8—3,3	2,0—2,8	6,0—8,0	0,1—0,5	0,2—0,8	230—250	160—180	1,5—2	445—540
AK5M2	4,0—6,0	1,3	1,5—3,5	0,2—0,8	0,2—0,8	200—220	120—150	1—2	410—570
AK5M4	3,5—6,0	1,2	3,0—5,0	0,2—0,5	0,2—0,6	200—230	140—180	0—1	380—440

Alloy	σ_u	$\sigma_{0,2}$	δ, %	K_{Ic}, N/mm³/²	H	F
	N/mm²				mm	
	T6					
Ak2Zh2	280—300	200—240	4—10	635—790	17,5	300—330
Ak2Zh2M2	330—380	280—300	2—4	540—700	12,5	330—380
AK2Zh2M7	340—370	300—330	1—1,5	440—600	10—12,5	350—400
AK5M2	260—320	250—300	0,5—1,5	410—635	10—12,5	340—380
AK5M4	280—320	280—320	0—1	350—410	10—12,5	350—400

*Remainder is Al.

Various methods of enhancing the plasticity and ductility of silumins and other cast aluminum alloys are known. They consist in reducing the volume fraction or the grinding and spheroidization of excess-phase particles of crystallization origin due to a change in the composition of the alloys or production procedures.

A certain increase in the plasticity and ductility of the alloys is achieved when their impurity content is lowered. This reduces primarily to limitation of the iron content for primary silumins, and to restriction of the content of iron, nickel, magnesium, and low-melting-point metals for secondary silumins, and the iron and silicon content in alloys of other systems. This means has been known and widely used for some time. An example is silumin AK7pch (AL9-1) in which the iron concentration is reduced to 0.3-0.5% instead of 0.6-1.5% in alloy AK7ch (AL9). In the strongest and most ductile alloys AK8M3ch (VAL8), AM4.5Kd (VAL10), and VAL12, the iron (and silicon) content is also limited and correspond to 0.2-0.4%.

High purity requires the use of pure grades of aluminum in smelting alloys and strict observance of casting technology. Under actual conditions, it is difficult to carry out these measures, while they are still impossible for secondary alloys.

It is not a reduction in the amount of excess phases of crystallization origin, but targeted variation in their morphology by alloying and changes in procedure that is the more promising method. A classical example is the modification of silumins, which ensures dispersion of a eutectic-colony structure [2]. The modification of silumins, especially copper-free ones, ensures improvement in their plasticity indicators. In copper silumins with a high iron content, however, modification of the structure of the silicon eutectic is not so effective. Here, modification is required to change the morphology of the iron-containing phases.

If these phases are crystallized in the form of spherical particles, they do not exert a negative influence on the plasticity and crack-resistance indicators. For example, additives of beryllium, which bonds the iron in the $Fe_2Al_4Be_5$ phase with a globular morphology, have been introduced to an $Al-7\%$ $Mg-3\%$ Zn alloy. As a result, that level of crack resistance that the alloy possessed without the iron additive was essentially completely restored (Fig. 1).

It is possible to neutralize the negative influence of silicon impurities in alloys based on $Al-Mg$, $Al-Mg-Zn$, and $Al-Zn-Mg-Cu$ systems in a similar manner, if formations of intermetallides of a denser form than Mg_2Si are obtained.

It is demonstrated in [3-5] that use of Be, Ni, and Co additives in iron-containing aluminum alloys, and alloying with alkali metals in silicon-containing alloys may result in the formation of globular phases. Appropriate phase diagrams, which have also been constructed [3-5], are required to determine the optimal concentration of additives in specific alloys.

The effectiveness of this method of improving the set of mechanical properties is illustrated by the results of investigation of an $Al-6\%$ $Zn-1.6\%$ $Mg-1\%$ $Cu-0.15\%$ Cr alloy with 0.5% Si, in which 0.2-0.5% of Ca was introduced. The intermetallide $CaSi_2Al_2$ is formed in this alloy during crystallization. Here, the magnesium is not consumed in bonding the silicon, and, remaining in the aluminum solid solution, ensures greater hardening of the alloy during aging: the yield point is increased from 400 to 450-460 N/mm², and the ultimate rupture strength from 470 to 500-510 N/mm². The relative elongation is simultaneously increased from 1 to 6-8%.
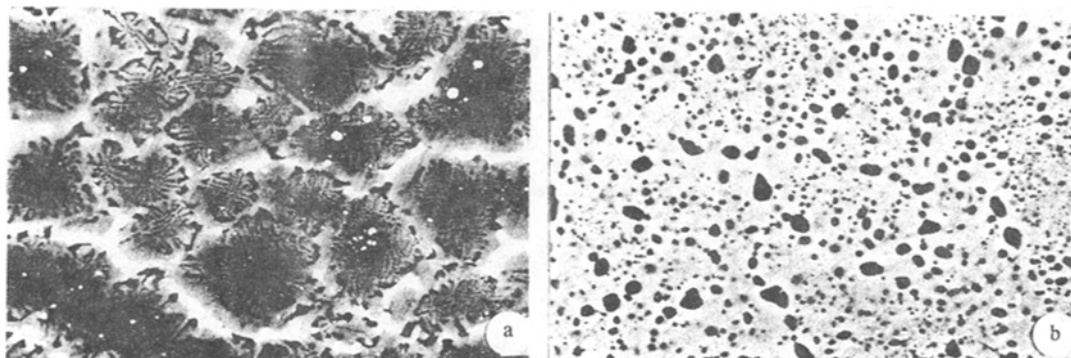
Fig. 3. Diagram showing distribution of regions of Fe- and Si-containing phases in Al−Fe−Si−(0-8%) Cu−(0-1%) Mg system at room temperature after crystallization at cooling rate of 10^2 °C/min. Phases initially crystallized are indicated in brackets.

In addition to phases of crystallization origin, the composition and structure of the crystals of the aluminum solid solution exert a significant influence on the plasticity and ductility of cast aluminum alloys. Copper and magnesium play the most important role in this in silumins. Their influence is associated with several factors simultaneously: the extent to which the aluminum solid solution is alloyed, the formation of additional excess phases (θ, W, Mg_2Si, etc.), coarsening of the silicon eutectic, and the appearance of secondary segregations during aging. On the whole, the elements in question, being basic hardners in silumins, exert a negative influence on their plasticity in any state. Here, the extent to which the aluminum solid solution is alloyed, especially with low copper (up to 1-2%) and magnesium (up to 0.2-0.3%) concentrations, probably lowers the plasticity to a greater degree than the excess phases that are formed during crystallization.

The influence exerted by the extent to which the aluminum solid solution is alloyed on the fracture toughness K_{Ic} is manifested differently, since K_{Ic} characterizes not only the material's ability to deform plastically (in a local volume), but also the stress under which it begins, which is associated with the level of the yield point. Additives of 0.3% Mg and 3% of Cu in a binary alloy raise K_{Ic} appreciably. At the same time, other crack-resistance characteristics (critical crack opening, work of failure, etc.) may be reduced significantly simultaneously with an increase in K_{Ic}. This requires a broader approach in selecting a structural-strength criterion for silumins.

As we know, aging to improve strength properties leads to a reduction in the plasticity of silumins often below the level of the cast state. A few published data and our results of evaluation of fracture toughness also suggest a reduction in K_{Ic} after aging in accordance with regime T6.

Overaging regime (T7) is used to elevate the plasticity indicators of silumins; here, the strength properties are reduced. Data on the determination of K_{Ic} indicate, however, that an increase in fracture toughness cannot be attained by this method. This is most likely because crack development should be facilitated where secondary segregations of the $\beta''(Mg_2Si)$ and $\theta'(CuAl_2)$ phases are in acicular and platelet form. Available data on the structure of the matrix in silumins indicate that all basic phase-hardeners — θ', θ'', β'', β', and S' — have a similar morphology. Only the size of the inclusions, their density, and the pattern of distribution submit to regulation; this is insufficient for attainment of the desirable structure with a uniform distribution of spherical particles. Methods of forming a similar structure in silumins have yet to be developed.

Thermally hardened silicon-free alloys based on the Al−Cu, Al−Mg−Zn, and Al−Zn−Mg−Cu systems exhibit broader possibilities for altering the structure of the dendrites in the aluminum solid solution. General laws governing structural changes during quenching and aging are the same as those in wrought alloys of similar composition [6, 7]. These alloys have maximum plasticity and crack-resistance characteristics after quenching. These indicators are reduced after aging, especially when regimes ensuring maximum strength are used. The optimal combination of strength, plasticity, and crack resistance of the alloys is achieved either with zone (low-temperature, for example, natural), aging, or after a certain overaging.

In addition to aging, quenching preceded by homogenizing annealing also affects the plasticity and ductility characteristics of the alloys. It is known [8] that dissolution of the nonequilibrium excess of the phases of crystallization origin during the quench heat increases all plasticity and crack-resistance characteristics. A change in the morphology of excess-phase particles during high-temperature heating also exerts an influence on these characteristics; this will be discussed below.
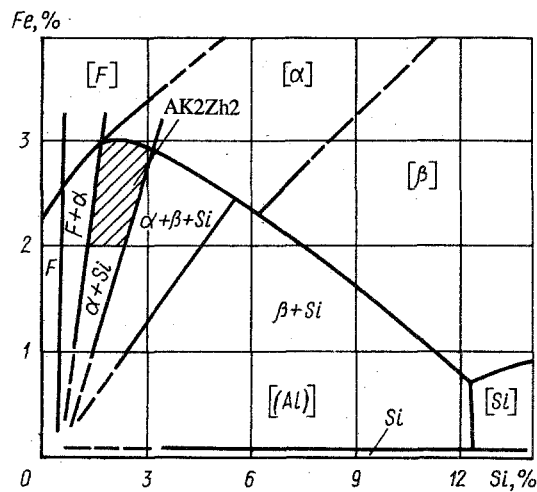
Fig. 2. Microstructure of Al + 6% Zn + 1% Cu + 8.5%
Mg + 4% Si alloy in cast state (a) and after holding at
460°C for 6 h + 510°C for h (b).

Among the other production parameters affecting the plasticity and ductility of castings, the cooling rate during crystallization should be mentioned first. Acceleration of cooling disperses the microstructure of the casting: the size of the dendritic cells of the aluminum solution and the particles of the excess phases of crystallization origin decreases. Such a dispersion of the microstructure always gives rise to an increase in plasticity. This effect has been known for some time [9] and is widely utilized. It is precisely the conversion to casting in metallic molds, for example, under pressure, that is therefore a basic trend in foundry development.

Particularly high mechanical properties of castings are attained when they are crystallized under pressure, which basically eliminates porosity [10]. During crystallization under pressure in a metallic mold, moreover, additional acceleration of cooling and dispersion of the microstructure occur due to elimination of the gap between the mold and casting.

Porosity in castings can be eliminated by applying hot isostatic pressure (HIP) to them at high temperatures in the solid state [11]. Such high-pressure treatment in gas cabinets ensures appreciable enhancement of all mechanical properties, and primarily the plasticity indicators of the alloys. After HIP at 460°C for 1 h under a pressure of 300 MPa, for example, the relative elongation of a type VAL12 alloy attains values of 8-17 instead of 3-7% after conventional heat treatment.

Use of the above-enumerated methods of affecting the structure of aluminum alloys makes it possible, even today, to obtain castings that are essentially superior to the deformed semifinished products in terms of a set of mechanical properties. Use of the methods under consideration is not always possible, however, under actual industrial conditions. For example, HIP for castings formed from aluminum alloys is almost never used due to the lack and high cost of gas cabinets; during the crystallization of alloys—nonsilumins in metallic molds, it is not at all possible to produce all castings without defects, and so forth.

As a result, improvement of the casting properties of alloys—silumins exhibiting high mechanical properties, but having inadequate fluidity and high hot shortness still remains an urgent problem. The basic trend in the solution of this problem is the introduction of additives that are slightly soluble in aluminum to these alloys to increase the volume fraction of eutectic components. A small amount of additives is introduced mostly for this purpose, since the mechanical properties are sharply reduced by significant alloying as a result of an increase in the fraction of eutectic components. The level of casting properties is insufficient, however, when a small amount of eutectic is present in the structure of the alloy.

In recent years, the alloying of aluminum alloys not with small additives, but with that amount of components that effect the formation of a large — as in silumins — volume fraction of eutectic (up to 20% and more), which consists of an aluminum solid solution and intermediate phases, has been proposed for realization of this trend. These phases should satisfy the following requirements:

1) result in the formation of disperse eutectic mixtures (this ensures good plasticity even in the cast state); and,

2) exhibit the ability to fragment and become spheroidized under high-temperature quench heat (this promotes the attainment of high plasticity and crack resistance in the heat-treated castings). The results of our investigations of alloys of the Al−Zn−Mg−Cu−Ni [12] and Al−Mg−Si systems can be given as an example of the implementation of this method.

A eutectic of α-solid solution + NiAl$_3$ with a disperse structure is formed when nickel is introduced to familiar high-strength alloys of the Al−Zn−Mg−Cu system. Here, the crystallization interval (for an Ni content of more than 4%) is constricted by more than 20°C, and the casting properties of the alloy are considerably improved. The strength properties of the alloy are also elevated after quenching and aging. Where $\sigma_{0.2}$ =425 N/mm^2, σ_u = 445 N/mm^2, δ = 5%, and K$_{Ic}$ = 1585 N/mm$^{3/2}$ in the case of the initial Al + 6% Zn + 1.6% Mg + 1% Cu alloy, this alloy possesses $\sigma_{0.2}$ = 480 N/mm^2, σ_u = 525 N/mm^2, δ = 3%, and K$_{Ic}$ = 950 N/mm$^{3/3}$ after alloying with 4% of Ni. Owing to the dense form of the NiAl$_3$ particles that form due to fragmentation of the eutectic crystals and their subsequent spheroidization, the plasticity and crack resistance of the alloy are not reduced very appreciably.

Alloy AMg4, 5K1.5 of the Al−Mg−Si system, which is employed in the automobile industry, contains 4.5-5.2% of Mg and 1.3-1.7% of Si; this promoted the formation of a considerable amount of the eutectic component into which the Mg$_2$Si phase enters. On heating above 500-520°C, the crystals of this phase, which have a complex nonequiaxial shape in the eutectic colony, fragment into individual inclusions and acquire a dense form during subsequent annealing.

Similar results are also obtained in alloys based on the Al−Mg−Zn and Al−Zn−Mg−Cu systems with a silicon additive (Fig. 2).

The α (Fe$_2$SiAl$_8$)-phase, which enters into the composition of the eutectic that may form in alloys based on different systems (Al−Cu, Al−Mg−Zn, Al−Zn−Mg−Cu) on adding 2-3% each of Fe and Si, exhibits properties similar to the intermetallide Mg$_2$Si. A high level of mechanical properties, including plasticity and ductility indicators amounting to 80-90% of the values of the initial characteristics with a simultaneous increase in casting properties can be retained for this type of developed alloys.

The α (Fe$_2$SiAl$_8$) phase can also be used in developing the composition of alloys that can be smelted from a mixed secondary raw material. We investigated the phase regions in an Al−Si−Fe−Cu−Mg−Mn system containing all the basic components of a secondary raw material in the actual range of their concentrations: up to 13% of Si, 4% of Fe, 10% of Cu, and 2% each of Mg and Mn. A diagram showing the distribution of phase regions for alloys in cast state is presented in Fig. 3. The content of copper and magnesium exerts a weak influence on the position of the phase boundaries, whereas even small changes in the concentrations of iron and silicon result in its pronounced variations. This phase diagram differs significantly from the equilibrium diagram as a result of suppression of peritectic reactions during accelerated cooling, the following reaction in particular: L + α → (Al) + β. As a result, the α-phase of eutectic origin remains in the cast structure and no β-phase needles are formed. It is established, moreover, that branched single α-phase crystals inside the eutectic colony are fragmented during high-temperature heating (for quenching) and tend to be spheroidized in a manner similar to the Mg$_2$Si phase (see Fig. 2). A structure, which is optimal in terms of morphology and which corresponds to maximum plasticity and crack resistance, is therefore formed.

As is apparent from Fig. 3, the phase region with a promising structure (cross-hatched) is located at concentrations of 2-3% of Fe and 2-3% of Si, which have never been previously employed for cast and wrought alloys. A series of new secondary alloys with mechanical and casting properties on an even higher level than those for familiar copper silumins could be developed with this combination of iron and silicon and a different ratio of magnesium, copper, and manganese (Table 2). The advantage of these alloys, however, consists not only in the properties, but primarily in the low cost and availability of the secondary raw material from which they can be smelted. These alloys have already been approved for industrial conditions and have been especially highly recommended for pressure casting — with a large amount of iron, the casting does not adhere to the mold.

REFERENCES

1. I. I. Novikov, Hot Shortness of Nonferrous Metals and Alloys [in Russian], Nauka, Moscow (1966).
2. Yu. N. Taran and V. I. Mazur, Structure of Eutectic Alloys [in Russian], Metallurgiya, Moscow (1978).
3. N. A. Belov, Yu. V. Evseev, and V. S. Zolotorevskii, "Relation between fracture toughness of cast Al−Mg−Zn alloy and morphology of excess phases," Izv. Akad. Nauk SSSR, Met., No. 6, 151-157 (1985).
4. V. S. Zolotorevskii, N. A. Belov, and T. A. Kurdyumova, "Optimization of the structure of secondary silumins to improve their plasticity and fracture toughness," Izv. Vyssch. Uchebn. Zaved., Tsvetn. Metall., No. 1, 76-88 (1989).
5. A. A. Aksenov, N. N. Belov, V. S. Zolotorevskii et al., "Microalloying of high-strength cast aluminum alloys with an elevated iron and silicon content," Izv. Akad. Nauk SSSR, Met., No. 1, 114-120 (1988).

6. V. G. Kudryashov and V. I. Smolentsev, Fracture Toughness of Aluminum Alloys [in Russian], Metallurgiya, Moscow (1976).

7. P. G. Miklyaev, G. S. Neshpor, and V. G. Kudryashov, Fracture Kinetics [in Russian], Metallurgiya, Moscow (1979).

8. V. S. Zolotorevskii, Structure and Strength of Cast Aluminum Alloys [in Russian], Metallurgiya, Moscow (1981).

9. I. F. Kolobnev, V. V. Krymov, and A. V. Mel'nikov, Nonferrous Casting of Lightweight Alloys. Foundryman's Handbook [in Russian], Mashinostroenie, Moscow (1974).

10. A. I. Batyshev, Crystallization of Metals and Alloys Under Pressure [in Russian], Metallurgiya, Moscow (1977).

11. N. S. Postnikov, Hardening of Aluminum Alloys and Castings [in Russian], Metallurgiya, Moscow (1983).

12. N. A. Belov, V. S. Zolotorevskii, and É. É. Tagiev, "Effect of nickel aluminide and magnesium silicide on the structure and mechanical and casting properties of the Al−Zn−Mg−Cu alloy," Izv. Akad. Nauk SSSR, Met., No. 1, 146-151 (1992).