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STRUCTURE AND PROPERTIES OF SILUMIN CASTINGS SOLIDIFIED UNDER PRESSURE AFTER HEAT TREATMENT

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Mechanical properties and structure of castings from hypoeutectic not modified silumin AlSi7Mg0.3 obtained by casting with crystallization under pressure and treated in a mode close to T6 are studied. The effect of the morphology of the particles of eutectic silicon on the properties of the alloy is determined. The influence of the casting modes and of the duration of holds of nozzle castings in the heating furnace before quenching on the properties of the silumin is described.

Key words: aluminum alloys, solidification under pressure, microstructure, spheroidization of eutectic silicon, mechanical properties.

INTRODUCTION

For replacing iron-base alloys with alloys based on light metals in the production of critical parts, the candidate metal should possess the specified minimum ductility in addition to the required strength. Specifically, its elongation should be no lower than 15%. In aluminum alloys AlSi7Mg0.3 and AlSi7Mg0.6 widely used in the automotive industry the attainment of such elongation (at the required strength) is problematic. This is explainable by the presence of coarse and brittle plates of eutectic silicon in the structure of the alloys, the shape of which is improved by inoculation. However, this does not always provide the required minimum permissible elongation.

It is known [1], that when a hypoeutectic silumin is heated above 500°C, the particles of eutectic silicon in it are spheroidized, and this may promote optimization of the mechanical properties of the alloy. The spheroidization manifests itself at a high fineness of the primary structure, which can be affected by regulating the rate of cooling of the melt [2]. When the heat transfer factor on the “solidifying casting – mold” interface is increased under the action of a high mechanical pressure under the conditions of casting with crystallization under pressure (PCC), the cooling rate can attain $10^2 – 10^3$ K/sec [3, 4]. This promotes formation of very fine particles of eutectic silicon.

In [5, 6] a favorable morphology of eutectic silicon (fibers of division *6* of the six-stage scale of Chai–Bäckerud [8]) has been obtained in cast condition. As a result of subsequent spheroidization of silicon, which develops at the start of a hold at the temperature of heating for quenching (above 500°C), an elongation $\delta = 15\%$ has been obtained after aging. Such heat treatment of silumins is known as silicon spheroidization treatment (SST). It is virtually identical to treatment modes T6 and T4 (quenching and artificial or natural aging), but differs from them by shorter holds at the temperature of heating for quenching.

At specific parameters of the PCC process with forced flow of the crystallizing melt [4, 7], which yields a nondendritic structure in the α-solid solution, we can obtain eutectic silicon with different shapes of crystals in accordance with the six-stage scale of Chai–Bäckerud [8].

The aim of the present work was to choose a method of crystallization and heat treatment modes for castings from not inoculated alloy AlSi7Mg0.3 for raising its elongation.

METHODS OF STUDY

Nozzle specimens (Fig. 1) were obtained from alloy AlSi7Mg0.3 by punch PCC, in which the final contours of the casting are formed by introduction of a pressing punch into the melt cast into a press mold. The shape of the casting and the temperature regimes of the application of pressure are necessary conditions for the appearance of shear stresses in the solidifying melt, which guarantee formation of a spheroidal nondendritic morphology of grains of α-solid solution in the structure of the alloy. Flow can arise in the melt as a re-

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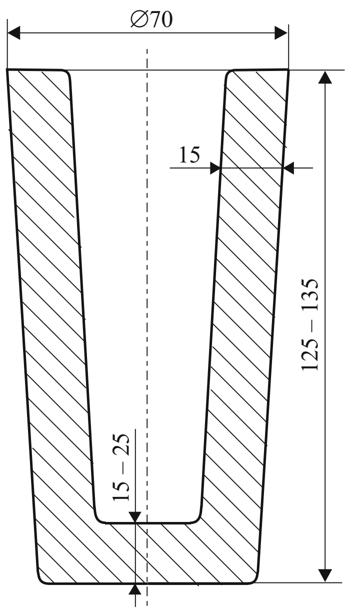


Fig. 1. A nozzle casting.

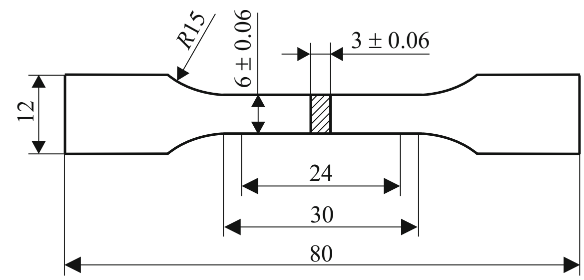


Fig. 2. A specimen for tensile tests.

sult of temperature and phase changes in the volume during cooling and solidification or as a result of disturbance of the mass balance of the solidifying alloy, when a part of it is removed from the cavity of the die. Both methods are described in [4, 7].

The castings for testing were fabricated by the method of PCC by the following variants: (*1*) flow of melt only due to volume changes in the process of cooling and solidification of the casting (pure punch pressing), (*2*) enhanced intensity of flow of not solidified alloy due to removing about 10% of its volume from the cavity of the mold during solidification, and (*3*) (a comparative mode) fabrication of the casting at a virtually atmospheric pressure; the contours of the casting were formed only under the action of the mass of the punch.

We used a PYE 250 SS M hydraulic press with rated force 400 kN and no-load speed 200 mm/sec; at the moment when the protruding part of the punch touched the melt poured into the mold, the speed was switched automatically to the working value equal to about 20 mm/sec. The temperature of the die of the press mold was stabilized in 90 min at 80°C. The punch was heated only by the thermal radiation from the die for 90 min, when the press mold was closed.

The material of the casting was aluminum alloy AlSi7Mg0.3 containing (in wt.%) 6.701 Si, 0.362 Mg, 0.112 Fe, 0.030 Ti, 0.015 Mn, the remainder Al. The composition of the alloy was close to that of the AK7ch counterpart (GOST 1583–93).

The alloy was melted under a layer of a SYLUKRIT flux; prior to filling the mold the metal was deoxidized at 720°C with an ECOSAL-AL 114 agent. The temperature of the alloy in the furnace before casting was controlled with the help of an OMEGA HH 306 OMEGATTE® temperature detector. The casting temperature in PCC is 615°C; the temperature of chill casting is 625°C.

When the temperature of the melt in the crucible attained 620°C, the press mold was opened, and the punch was lifted to the initial position. When the temperature of the melt decreased to 617°C, the melt was poured into the die. Five seconds after the start of the pouring the press mold was closed by installing the punch into the poured melt at a pressing pressure of about 100 MPa. After a hold at this pressure for 30 – 32 sec, the pressing was stopped, and the casting was withdrawn from the die of the press mold.

The following parameters of the PCC process were recorded: the temperature of the solidifying casting, the temperature of the press mold, the displacement of the pressing punch, and the pressing pressure (in the hydraulic system of the press). The accuracy of the temperature measurement was ± 0.1°C; that of the pressure measurement was ± 0.1 MPa, that of the displacement of the punch was ± 0.01 mm. All the parameters were recorded by a computer at a frequency of 0.01 sec.

The temperature in the furnace before the heat treatment of the castings by regime T6 was stabilized for 6 h at 540°C. The hold of the casting at this temperature lasted for 3, 5 or 10 min as well as in the SST (in what follows we will denote this mode of treatment T6×3, T6×5 and T6×10, respectively) [5, 6]. The time of attainment of a temperature ≥ 500°C in the center of a specimen had been determined earlier (2 min). The time (heating + holding) of residence of the castings (specimens) in the furnace was 5, 7, and 12 min, after which they were quenched in water with a temperature of 20°C. Artificial aging at 160°C for 4 h was conducted in the furnace, which was heated preliminarily to the specified temperature. After the artificial aging the specimens were cooled in air.

The mechanical characteristics of the alloy were determined by tensile tests of the specimens (Fig. 2) in a universal LabTest 5.250 SP1 machine (at a force of 250 kN) moving the cross member at a speed of 2.5 mm/min. The working diagram recorded by the computer was used to compute the rupture strength σ_r, the yield strength $\sigma_{0.2}$, and the elongation δ.

The microstructure of the castings was studied with the help of a light microscope (Zeiss Axiovert 40 MAT) and a scanning electron microscope (JEOL JSM-6610).

TABLE 1. Mechanical Properties of Castings from Silumin AlSi7Mg0.3 Produced by the Method of PCC

Casting variant	σ_r, MPa				$\sigma_{0.2}$, MPa				δ, %			
	F	T6×3	T6×5	T6×10	*F*	T6×3	T6×5	T6×10	*F*	T6×3	T6×5	T6×10
	Silumin AlSi7Mg0.3											
1	315	374	379	371	133	154	156	152	11.0	18	18	21
2	314	367	386	381	152	168	174	178	12.0	16	20	15
3	234	298	297	324	135	144	157	164	4.5	6	6	8
	Silumin A356-T6 (AlSi7Mg0.3) [10]											
PCC	–		300		–		225		–		12	

Notations: *F*) the properties of alloy AlSi7Mg0.3 prior to heat treatment; T6×3, T6×5, T6×10) after heat treatment involving a hold of the specimen at 540°C for 3, 5, and 10 min, respectively, followed by water quenching and 4-h aging at 160°C.

RESULTS AND DISCUSSION

We established that the rate of cooling during solidification of the central zone of the vertical wall of the casting (at equal distances from the ends) in the process of PCC was $(6 - 8) \times 10^2$ K/sec, which promoted formation of a primary fine-grained structure.

Mechanical Properties. The characteristics of the mechanical properties of the alloy are presented in Table 1 (average values after testing three specimens). For comparison, we also present the properties of PCC-castings from alloy A356 (AlSi7Mg0.3) after treatment T6 of the NADCA Standard [10]. The properties of the castings before the heat treatment are marked by letter *F*; those due to the treatment are marked T6×3, T6×5 and T6×10. It can be seen from Table 1 that the strength characteristics of the castings produced by variants *1* and *2* of PCC in cast condition are higher than those of the standard PCC castings from alloy A356 after a heat treatment by variant T6 [10]. The solid solution is supersaturated due to the high rate of cooling of the castings in the liquid and solid conditions (in fact, after quenching from liquid state [9]) and due to lowering of the diffusion in the liquid condition as a result of the action of pressure on the cooling and crystallizing alloy. It should be noted that the castings (specimens) have been subjected to natural aging for several weeks before the heat treatment and testing for the tensile strength. This seems to be close to the condition of castings heat treated by regime T5. In our case the quenching was replaced by rapid cooling from liquid state under mechanical pressure, and the aging was artificial.

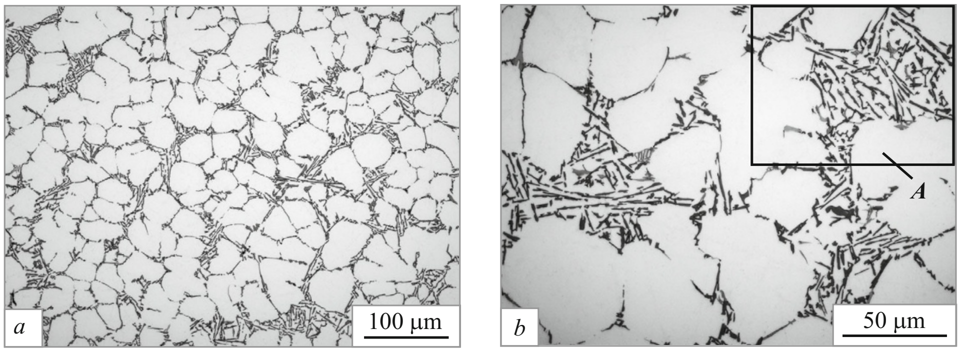
Note that after variant T6×3, i.e., after a 3-min hold of the casting at the temperature of heating for quenching, the maximum value of the ultimate rupture strength $\sigma_r = 370$ MPa, which is 50 – 60 MPa higher than in the alloy in cast condition. A similar change is typical for $\sigma_{0.2}$ too. The elongation δ increases from 11 to 18% (casting variant *1*) and from 12 to 16% (casting variant *2*), and the possibility of its further growth is not exhausted (as compared to regimes T6×5 and T6×10).

Prolongation of the hold time at the temperature of heating for quenching to 5 min has shown that the alloy cast by variants *1* and *2* has $\sigma_r = 379$ and 386 MPa respectively. The yield strength $\sigma_{0.2} = 156$ MPa (variant *1* and treatment T6×5) and $\sigma_{0.2} = 178$ MPa (variant *2* and treatment T6×10). The holds at the temperature of heating for quenching longer than 10 min lower the value of σ_r.

The elongation is more sensitive to variation of the morphology of silicon particles in the castings, including those obtained with forced flow of the melt. After casting by variants *1* and *2* the average values of δ = 11 and 12%, respectively, which is close to the elongation of alloy A356 cast by the method of PCC [10]. A maximum value of δ = 21% has been obtained after casting by variant *1* and heat treatment T6×10; δ = 20% has been obtained after casting by variant *2* and treatment T6×5.

With allowance for the attained values of σ_r and $\sigma_{0.2}$ the preferable modes of production are casting by variant *2* and heat treatment T6×5. The castings obtained by variant *3* do not meet the principal requirement of minimum value of δ = 15%. The strength characteristics of such castings are also lower than those of the PCC ones.

Microstructure. With allowance for the results of the mechanical tests we studied the microstructure of castings produced by variant *2*. Some of them are presented in Figs. 3 – 5. Figure 3*a* shows nondendritic (spherical) grains of the α-solid solution surrounded chiefly by fine plates of eutectic silicon. Such a structure forms under a high rate of cooling of the casting during solidification (6×10^2 K/sec). The morphology of such eutectic silicon can be described by divisions 2 – 3 of the six-stage scale of Chai–Bäckerud [8]. The formation of the fine structure is largely responsible for the high values of strength σ_r and ductility δ of the castings. The length of the silicon plates in the plane of the lap differs. The structure contains plates of eutectic silicon of two kinds (Figs. 3*b* and 4*a*), i.e., short plates with sharp protrusion (*1*) and relatively long and thin plates (*2*). At a high magnification the sharp protrusions are well observable (Fig. 3*b*). In



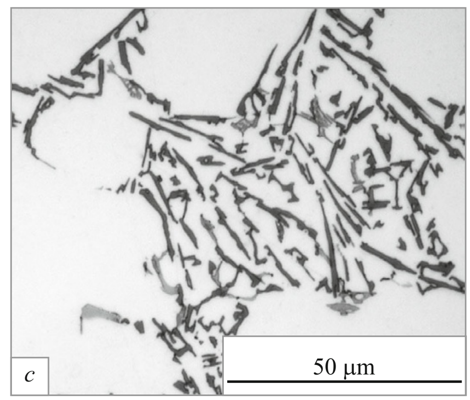


Fig. 3. Microstructure of a casting from silumin AlSi7Mg0.3 produced by variant *2*: *a*) solid solution of a nondendritic quasi-spheroidal morphology; *b*) plates of eutectic silicon: short ones with sharp protrusions (region *A*) and relatively long and thin ones (in the middle of the height of the image); *c*) enlarged region *A*.

addition to the thin and short silicon plates with an average thickness of 1 μm (Fig. 4*a*) the structure contains long plates of eutectic silicon that are a feature of the start of spheroidization (Fig. 3*b*). A more detailed study has shown that all the particles of eutectic silicon in cast condition have sharp faces. The presence of a phase with such morphology raises the internal energy of the system.

Heating to 540°C with a hold of 3 min promotes formation of a state with lower surface energy. As a result, the eutectic silicon looses the sharpness of the edges; the edges and protrusions of plates or coarse fibers are gradually dissolved and rounded (Fig. 5*a* and *b*). This is also confirmed by the results of the study of the structure under a scanning electron microscope (Fig. 4*b*).

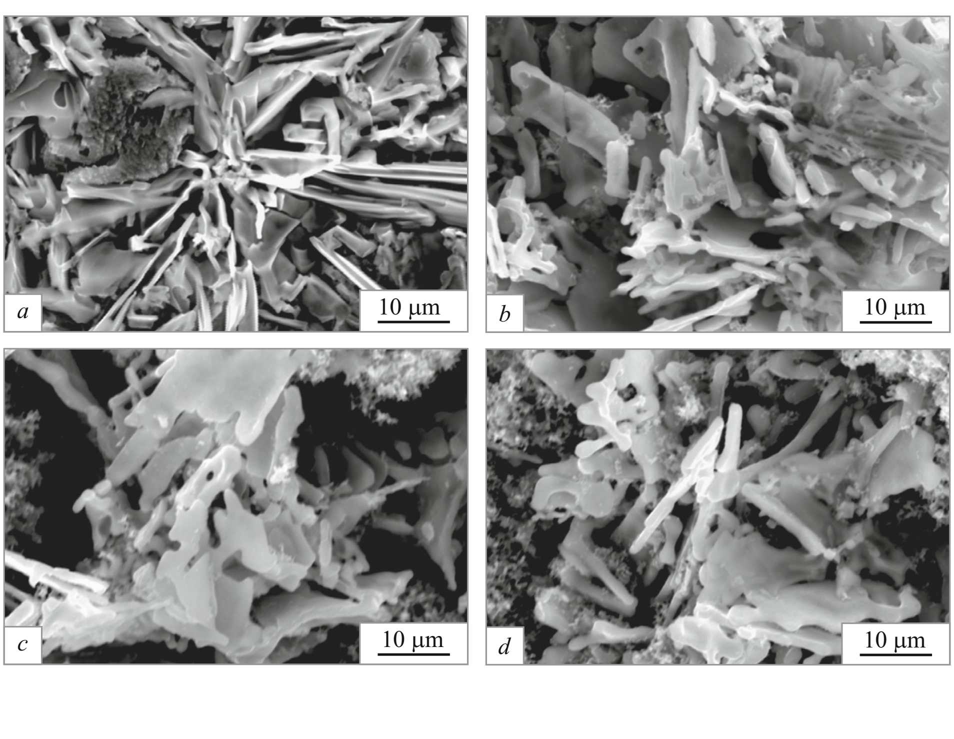


Fig. 4. Eutectic silicon in the structure of alloy AlSi7Mg0.3 (scanning electron microscopy): *a*) after casting by variant *1*; *b*, *c*, *d*) after casting by variant *1* and heat treatment by regimes T6×3, T6×5, and T6×10, respectively.

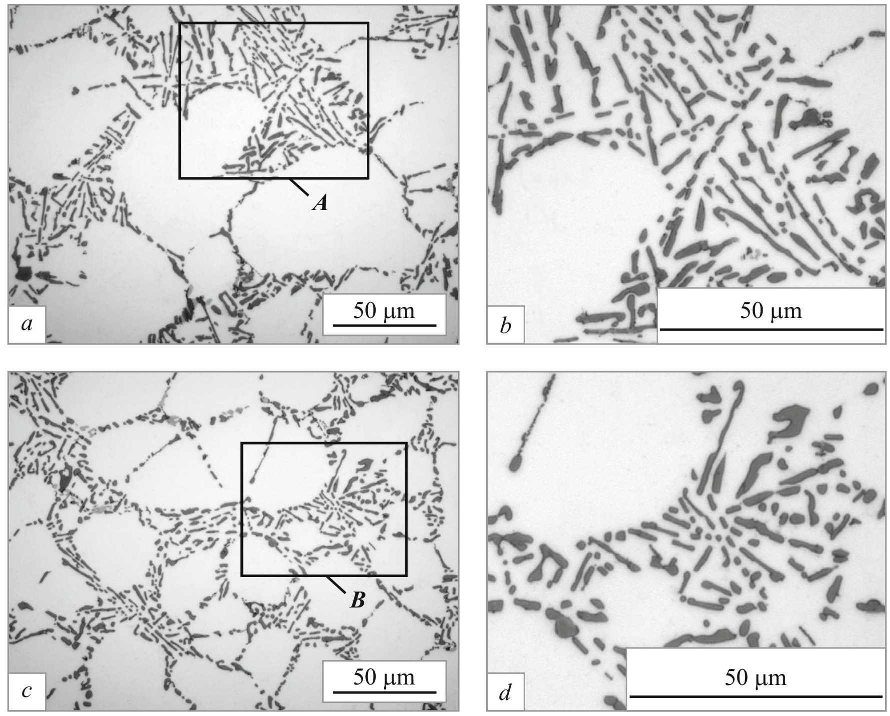


Fig. 5. Microstructure of silumin AlSi7Mg0.3 after casting by variant *2* and heat treatment by regimes T6×3 (*a*, *b*) and T6×5 (*c*, *d*): *a*, *c*) general appearance; *b*, *d*) regions *A* and *B*, respectively, with rounding of the sharp edges of silicon plates.

When the time of the hold at 540°C is increased to 5 min, the process of rounding of the ends of the plates of eutectic silicon proceeds intensely (Fig. 4*c*). The length of the silicon plates in the plane of the lap decreases progressively, and the particles acquire a round shape (Fig. 5*c* and *d*). Formation of such particles of eutectic silicon should provide operation of the articles under high stresses and strains without disturbing their continuity. After a hold of 10 min (T6×10) the particles of eutectic silicon are rounded still more and become coarser (Figs. 4*d* and 6).

Figure 4*c* and *d* present a structure with rounded protrusions on the ends of the plates and spheroidal formations at the ends of coarse fibers connected with the original fibers through a thin bridge. Dissolution of these bonding bridges is a prerequisite for the appearance of individual spherical particles of eutectic silicon.

A metallographic analysis has shown fineness of the structure of PCC castings. As a result of the high rate of cooling in liquid condition we may expect some instability of the structure of the alloy and, as a consequence, its elevated sensitivity to heat treatment [9]. Published data and the results of the present study show that gradual degradation the sharpness of plates of eutectic silicon (the process of spheroidization) develops even in the early stages of holding at the temperature of heating for quenching. This promotes elevation of the mechanical properties (σ_r and δ).

Substantial changes in the morphology of eutectic silicon were detected when we compared the microstructures of the alloy in cast condition and after a treatment by regime T6×3. Such modes of treatment of castings provided the highest growth in the mechanical characteristics. For the castings obtained by variant *2* maximum values of the mechanical characteristics were obtained after the heat treatment by regime T6×5. Subsequent prolongation of the hold at the temperature of heating for quenching caused coarsening of the plates of eutectic silicon and lowering of the ultimate rupture strength and of the elongation.

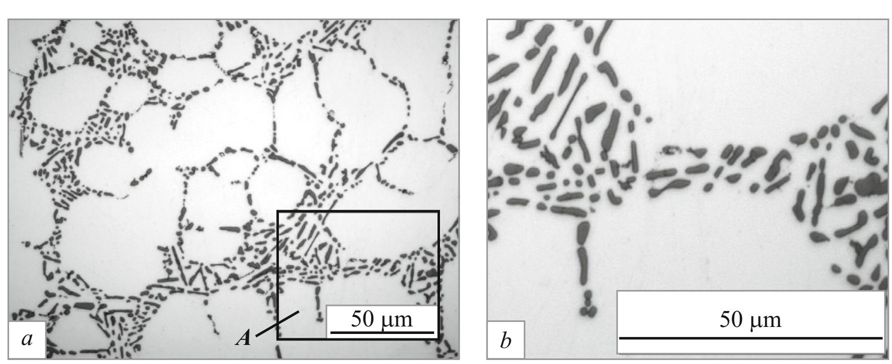


Fig. 6. Microstructure of silumin AlSi7Mg0.3 after casting by variant *2* and heat treatment by regime T6×10 (*b*): *a*) general appearance; *b*) region *A* with rounding of the sharp edges of silicon plates.

CONCLUSIONS

1. A high rate of cooling (6×10^2 K/sec) of alloy AlSi7Mg0.3 right after application of pressure (about 100 MPa) during solidification of a casting with forced flow of the melt provides substantial refinement of particles of eutectic silicon (to 2 – 3 divisions of the six-stage scale of Chai–Bäckerud).

2. The initial structure of castings should be fine-grained and metastable for subsequent spheroidization of eutectic silicon in heat treatment and growth of the ultimate ruptures strength and elongation (to 15%). In addition, the structure should possess a high sensitivity to the heat treatment used.

3. The most effective mode of heat treatment for castings from a not inoculated alloy AlSi7Mg0.3 produced with the use of PCC is T6×5, i.e., 2-min heating to 540°C, a hold for 5 min, cooling in water at 20°C and 4-h aging at 160°C.

4. The possibility of the use of a wide range of holds (for 3, 5 and 10 min) at the temperature of heating for quenching makes it possible to apply the SST treatment to castings with different wall thickness.

5. The spheroidizing heat treatment not only raises the mechanical properties of the castings but gives economic advantages due to the possibility of shortening the hold at the temperature of heating for quenching.

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