

UDC 669.715'721'884

PHASE COMPOSITION AND MECHANICAL PROPERTIES OF ALLOYS OF THE Al – Mg – Li – Me SYSTEM

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Translated from *Metallovedenie i Termicheskaya Obrabotka Metallov*, No. 2, pp. 40–43, February, 2008.

The phase composition and the structure of semiproducts from alloys of the Al – Mg – Li – Zr – Sc (1423) and Al – Mg – Li – Cu – Zr (1430) systems are studied with the use of x-raying. The temperature of dissolution of the ternary phase in heating of ingots from alloy 1423 is determined. The results of estimation of the structure-and-phase state and of mechanical properties are used to determine the factors required for raising the process ductility of sheets from alloy 1430.

INTRODUCTION

Reduced-density aluminum alloys of the Al – Mg – Li – Me system have a composite phase composition in the stages of intermediate and final heat treatment. This is a result of a considerable total content of alloying elements in the alloys as compared to the traditional alloy D16, which they are to replace. For example, the total content of alloying elements in alloy 1423 (the Al – Li – Mg – Zr – Sc system) amounts to 11.0 at.%; in alloy 1430 (the Al – Li – Mg – Cu – Zr system) the figure is 9.5 at.% [1]. Almost all of the alloying elements have variable solubility in aluminum. In this connection, the alloys bear ternary and binary intermetallic phases in addition to the main δ -Al₃Li hardening phase (Table 1). Segregation of these phases can occur in the process of hot deformation, homogenizing (in heating and cooling), and intermediate softening heat treatment of cold-deformed semiproducts.

A special feature of aluminum alloys of the Al – Mg – Li – Me system is their reduced ductility after annealing, which should cause softening in principle. However, these alloys in the annealed state possess substantial heterogeneity, i.e., particles of intermetallic phases are present over boundaries in the bulk of grains in an amount up to 6%. In addition, the main δ' -Al₃Li hardening phase segregates from the solid solution even in the process of delayed cooling. The location of phases (usually ternary ones) over grain boundaries is responsible for incompleteness of the recrystallization process in rolled semiproducts upon heating to a high temperature.

The aim of the present work consisted in substantiating the choice of heat treatment mode due to studying the phase

composition and the mechanical properties of ingots from alloy 1423 and hot-rolled sheets from alloy 1430.

METHODS OF STUDY

In order to study the phase composition of alloys of the Al – Mg – Li – Me system in the cast state and after homogenizing we used laboratory heats with the chemical composition presented in Table 2. The chemical composition of heat 8 corresponded to alloy 1423, that of heat 9 corresponded to alloy 1420, and that of heat 10 corresponded to alloy 1430. In the other heats we varied the contents of lithium and magnesium, and in some of them we added copper. In order to choose the appropriate mode for homogenizing annealing and to determine the temperature of the dissolution of the S₁-phase we performed quenching from 360, 450, and 490°C in water.

The structure of the intermetallic phases, the state of the solid solution, and the microstructure were studied on specimens cut from the middle of ingots.

TABLE 1. Structure and Parameters of the Crystal Lattice of Intermetallic Phases in Alloys of the Al – Mg – Li – Cu – Zr – Sc System

Phase	Crystal structure	Lattice parameters, nm
S ₁ (Al ₂ MgLi)	Cubic*	$a = 2.03$
δ' (Al ₃ Li)	fcc	$a = 0.403$
β' (Al ₃ Zr)	Tetragonal	$a = 0.4013, c = 0.1732$
β' (Al ₃ Sc)	fcc	$a = 0.4106$
T ₂ (Al ₆ CuLi)	Cubic**	$a = 0.139$

* 456 atoms in an elementary cell.

** 162 atoms in an elementary cell.

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TABLE 2. Chemical Composition of Tested Alloys

Heat	Content of elements, wt.%			
	Li	Mg	Zr	other
1	1.5	2.90	0.10	0.1 Sc
2	1.5	3.40	0.10	–
3	1.5	3.50	0.10	0.5 Cu
4	1.5	5.00	0.10	0.5 Cu
5	2.0	1.80	0.10	–
6	2.0	3.70	0.10	–
7	2.0	3.70	0.10	0.5 Cu
8 (alloy 1423)	2.0	3.50	0.08	0.15 Sc
9 (alloy 1420)	2.0	5.00	0.10	–
10 (alloy 1430)	1.66	2.77	0.11	1.40 Cu

Alloy 1430 was studied after hot rolling (sheets with a thickness of 4 mm) and after heating to a temperature of 420, 440, and 525°C with subsequent cooling both with the furnace and in water. After heat treatment in different modes we performed standard mechanical tests of the sheets and determined the ultimate rupture strength σ_r , the yield strength $\sigma_{0.2}$, and the elongation δ . We also computed the factor K that characterized the process ductility of the sheets upon subsequent cold deformation, i.e., $K = (\sigma_r / \sigma_{0.2}) \delta$.

Diffraction analysis was performed with the help of a DRON-3 device in copper K_α radiation. The diffractograms were deciphered using the ASTM data bank. The degree of development of the recrystallization processes was evaluated in terms of the results obtained with the use of an URK camera and an URS-2.0 device in cobalt and copper K_α radiation.

RESULTS AND DISCUSSION

Figure 1 presents a phase diagram of the Al – Mg – Li – Zr – Sc system plotted for a temperature of 450°C. After heating at 450°C only heats 1 and 2 with reduced content of lithium and magnesium (1.5% Li, 2.9 and 3.4% Mg, respectively) matched the single-phase region. All the other heats including those with composition corresponding to alloys 1420 and 1423 lay in the double-phase ($\alpha + S_1$) region at this temperature. The ternary S_1 phase after homogenizing was chiefly located over boundaries of dendrite cells in the form of a continuous or broken chain of particles 1 – 3 μm in size (Fig. 2). When the content of magnesium was decreased, the amount of the S_1 -phase on grain boundaries decreased too; the distribution of the particles became more uniform. The introduction of scandium (heat 1) did not affect the amount and distribution of the S_1 -phase.

Test hardening has shown that full homogenizing, i.e., dissolution of the S_1 -phase in alloy 1423, occurs after a hold at 490°C. However, homogenizing of the alloy should not be performed at 490°C, because full dissolution of the S_1 -phase causes subsequent decomposition of the solid solution in de-

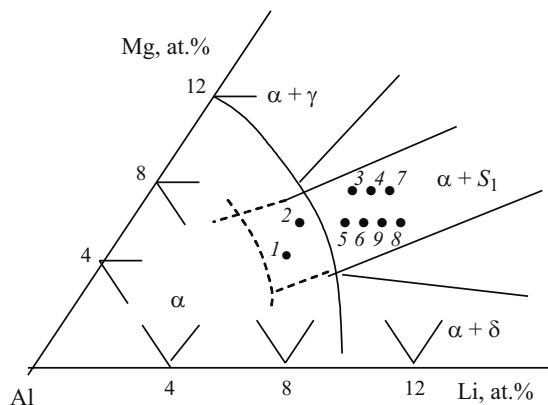


Fig. 1. An angle of a phase diagram of the Al – Mg – Li system plotted experimentally for alloys containing Zr and Sc at 450°C. The dots present the approximate location of the composition of tested heats on the diagram. The numbers at the dots denote the numbers of the heats.

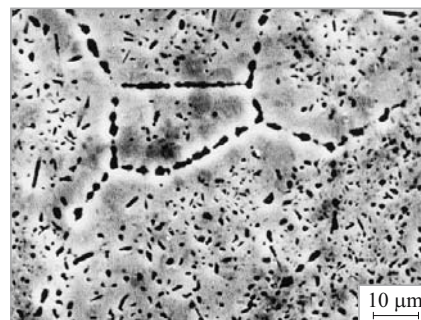


Fig. 2. Microstructure of an ingot from the metal of heat 8 after annealing at 450°C.

layered cooling. Segregation of the S_1 -phase depletes the solid solution of lithium, which lowers the efficiency of the subsequent aging (after the hardening). In this connection, annealing of alloy 1423 cannot be called a homogenizing process. The temperature of the annealing should correspond to some heterogeneity of the structure, i.e., the presence of particles of the S_1 -phase, which is ensured by heating to 450 – 460°C. The amount of the S_1 -phase in this case is 5 – 6%, which is 2 – 3 times lower than in alloy 1420 [1]. The introduction of copper (0.5%) lowers the temperature of the dissolution of the S_1 -phase to 450°C.

We studied the phase composition and the structure of alloy 1430 with the aim of determining the mode of heat treatment of hot-rolled sheets that would ensure maximum ductility in subsequent cold rolling.

Hot-rolled sheets had a partially recrystallized structure with nonuniform grain size (ranging from 1 to 5 μm), which contained intermetallic phases, i.e., a ternary T_2 (Al_6CuLi_3) phase and a β' (Al_3Zr) dispersoid.

Annealing at 420 and 440°C caused formation of a homogeneous recrystallized structure after cooling both with the furnace and in water; the degree of the recrystallization

TABLE 3. Mechanical Properties and Phase Composition of Hot-Rolled Sheets from Alloy 1430 after Heat Treatment

Heat treatment mode		Structure	Intermetallic phases	σ_r , MPa	$\sigma_{0.2}$, MPa	δ , %	K
t , °C	Cooling medium						
420	Furnace	Partially recrystallized	T_2, S_1, β' (Al_3Zr)	275	170	14	23.13
	Water			–	–	–	–
440	Furnace	Recrystallized		277	187	14	20.73
	Water			–	–	–	–
525	Furnace	Start of secondary recrystallization	$(S_1 + T_2)^*$, β' , δ'	257	133	16	30.9
	Water			$(S_1 + T_2)^{**}$, β'	354	190	23
530 → 480 °C (in air) → 100 °C (with furnace)		Staged mode		260	140	14	25.8

* The total amount is 4 vol.%.

** The total amount is 0.5 vol.%.

Note. Cooling conditions of staged mode given in parenthesis.

changed with the temperature but little. A ternary S_1 -phase formed in the alloy simultaneously. The amount of this intermetallic phase in alloy 1430 was about twice lower than in alloy 1423. This seems to be connected with the presence of copper in the alloy.

The mechanical properties of alloy 1430 after annealing at 420 and 440°C have close values (Table 3).

A hold at 525°C causes partial dissolution of phases T_2 and S_1 ; their total content after cooling in water decreases to 0.5%. Weakening of the decelerating effect of the grain-boundary segregations of the S_1 -phase due to migration of grain boundaries is responsible for the start of secondary recrystallization.

The rate of cooling from 525°C affects the phase composition and the properties of the sheets. As a result of cooling with the furnace the solid solution decomposes, yielding ternary phases in an amount up to 4% and a hardening δ' - Al_3Li phase. This is confirmed by the lower value of the lattice parameter of the solid solution as compared to that after water hardening and by the data of transmission electron microscopy.³ Particles of the δ' - Al_3Li phase 3–4 nm in size form uniformly in the bulk of grains. The structure becomes heterogeneous upon formation of intermetallic phases of a different nature over grain boundaries and in the bulk of grains, but this does not cause a substantial growth in the elongation and in the factor K that characterizes the process ductility of the alloy (Table 3).

Thus, the presence of recrystallized structure after annealing in multiphase alloy 1430 does not ensure the required growth in process ductility. In order to satisfy the requirements on this parameter we should eliminate or restrict the formation of the S_1 - and T_1 -phases in cooling. This is possible only if the cooling is conducted at a considerable

rate from 525°C. The use of water as a cooling medium promotes simultaneous growth in the strength and ductility parameters (Table 3). After hardening from 525°C the value of K is 25% and almost 50% higher than after annealing at 525 and 420°C, respectively.

An alternative treatment variant for sheets is the use of two-stage cooling (Table 3), i.e., an accelerated stage at 530–480°C (in air) and a slower stage at 480–100°C.

CONCLUSIONS

1. We have plotted a region of a ternary phase diagram and determined the temperature for homogenizing of ingots from alloy 1423.

2. Decrease in the content of magnesium from 5.0 to 3.5% produces a favorable effect on the structure of Al–Mg–Li–Me-alloys after homogenizing due to the reduction of grain-boundary segregations of the S_1 -phase in the ingots.

3. The process of recrystallization in heating of hot-rolled sheets from alloy 1430 is controlled by the amount and location of ternary S_1 - Al_2MgLi and T_2 - Al_6CuLi phases over boundaries and in the bulk of grains. In order to ensure maximum ductility in the sheets in subsequent cold rolling we should not only obtain a recrystallized structure but also prevent the decomposition of the solid solution that yields S_1 -, T_2 -, and δ' - Al_3Li phases. This can be achieved by accelerated cooling from 525°C.

REFERENCES

1. I. N. Fridlyander, K. V. Chuistov, A. L. Berezina, and N. I. Kolobnev, *Aluminum-Lithium Alloys. Structure and the Properties* [in Russian], Naukova Dumka, Kiev (1992),
2. I. N. Fridlyander, N. I. Kolobnev, L. B. Khokhlatova, et al., in: *Proc. 7th Int. Conf. on Aluminum Alloys (ICAA-7)* [in Russian], Vol. 3, Moscow (2000), pp. 1393–1397.

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