Fundamentals of Developing Compositions for Weldable Wrought Aluminium Alloys with Enhanced Heat Resistance

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Abstract

The paper presents in the form of diagrams the results of thorough research into fluidity and hot shortness susceptibility of the aluminium-rich area of the Al-Cu-Mg system carried out using the method of the mathematical planning of experiments (Scheffe simplex). A relationship is demonstrated between thermal stability of an alloy solid solution and the influence of Cu and Mg atoms vacancy structures, and diffusion processes in the solid solution. The dependence of heat resistance of aluminium alloys on the content ratio of Cu and Mg in them is analyzed. A number of characteristics values for aluminium alloys of the Al-Cu-Mg system with different Cu and Mg contents are compared. For the first time, the existence of a aluminium heat-resistant alloys that have good weldability has been substantiated experimentally and proved theoretically

1. Introduction

Among the known alloying systems, of special significance are the Al-Cu and Al-Cu-Mg systems, since only they form the basis for developing heat-resistant aluminium alloys. The main factor is the presence in the structure of stable strengthening phases: θ (Al₂Cu) in the Al-Cu system and S (Al₂CuMg) in the Al-Cu-Mg system (in the present paper the influence of supplementary alloying additives is not looked into).

Aluminium-copper alloys, as distinct from duralumins (AI-Cu-Mg), possess reduced values of the heat and corrosion resistance, somewhat higher density and tendency to heterogeneity of structure in semi-finished products (which decreases stability of properties), but, for all that, have the advantage when operating under the cryogenic temperatures conditions. Based on the AI-Cu system weldable heat-resistant aluminium alloys were developed for the first time: in the USSR – 1200 (D20), 1201, 1205[1-3]; in the USA – 2219, 2319, 2419; in the UK – 2021.

The Al-Cu-Mg system alloys possess higher heat resistance, since there is the S-phase in their structure [3, 4]. With a view to increasing heat and corrosion resistance of weldable wrought alloys in the Soviet Union, starting with the mid 1950s, researches were carried out on the Al-Cu-Mg system alloys [5, 6]. Attempts at welding the then existing duralumins produced negative results. Later on compositions were developed for the new weldable alloys of this system: M40 (1171) and, then, VAD1 (1190). After thorough investigations and commercial mastering, the alloys were applied in welded structures of aircrafts.

The enhanced heat resistance of these alloys is the result of high saturation with Mg (especially alloy M40) and, as a consequence, the position in the α +S phase area. However, these alloys did not find wide application in welded ware, since during their development there were no special investigations conducted into weldability, and production facilities were not completely prepared for the peculiarities of welding such alloys. It became necessary to perform in-depth investigations into weldability of the Al-Cu-Mg system aluminium alloys to choose optimum compositions for the new alloys.

2. Results and Discussion

Seeing that the basic criteria of weldability are fluidity and hot-cracking susceptibility, a systematic comprehensive study was performed of change in the mentioned characteristics of the Al-Cu-Mg system, as dependent on the Cu and Mg contents in the alloys and on the peculiarity of phase transformations in the temperature range of crystallization. Boundaries (according to the Mg and Cu percentages) of the area investigated encompassed all the main commercial duralimins and probable compositions of new alloys. Investigations were carried out using the aluminium alloys which contained, apart from Cu and Mg, alloying additives and impurities that are present in commercial alloys. Methodogically the operation was performed with the use of formal experiment planning (Scheffe's simplex-plans) and with mathematical treatment of the data obtained. The results of investigations are set forth in detail and analyzed in papers [7, 9], graphically they are presented on Figure 1, 2.

The pattern of hot cracking and the distribution of fluidity values explains the lack of weldability in existing duralumins; their compositions (mass) – 0.8-6.0% Cu; 0.2-1.5% Mg are positioned in the area (Figure 1-4) with the highest values of hot cracking susceptibility (30-40mm) and the lowest of fluidity (130-270mm). The stated peculiarity of the alloys can also exert a negative influence on the ingot quality and cause spoilage during casting. The analysis of the results of the investigations explains why the increase in Mg content (>0.02% mass) in alloys of the 2219 type leads to drastic growth of hot cracks during fusion welding; it also clarifies why problems arise during welding of alloys M40 (1171) and VAD (1190). In all the cases mentioned the chemical compositions of alloys, by Cu and Mg (Figure1), are in the area with the enhanced effective crystallization intervals, with the presence of critical fluid volume at the last stage of crystallization, and, as a consequence, with increased hot cracking.

The major outcomes of the long-term thorough investigations into the influence of Cu and Mg on the hot-cracking susceptibility and fluidity has been the development of theoretical and practical proof of the possibility of creating weldable alloys within the Al-Cu-Mg system, establishing the Cu and Mg concentrations which ensure the best combinations of hot cracking and fluidity values. Further research was directed towards studying the factors governing the influence Cu and Mg on the heat resistance of duralumins [13]. Analysis of the structure and properties of some of the domestic alloys of the Al-Cu-Mg system, as well as comprehensive investigations of the alloys of the shaded area of Figure 4 showed that the strength and heat resistance depend on the content ratio of the θ and S phases in the alloy structure. For example, the S-phase fraction increases as Mg content rises and, as a consequence, heat resistance increases [1-4, 14-21].

It is a well-known fact that many structural transformations in alloys (including the formation and growth of small particles) are a consequence of diffusion processes that are

closely connected with lattice defects and with interatomic interactions between the solvent and solute.

The availability and migration of vacancies facilitate the phase stages of ageing. Thus, in the alloy of the RR58 type [22] nucleation of the S' precipitation occurs at helicoids and loops formed by mono- and polyvacancies.

The presence of alloying element atoms (impurities) in aluminium influences the migration of vacancies, since the migration energy of vacancies in aluminium is quite similar to their bonding energy with atoms of some alloying elements. Drawing on, for example, reference [19], (including reference to a number of authors) the following data are adduced: the vacancy migration energy in pure aluminium is equal to 0.45-0.58 eV (aluminium self-diffusion energy 0.58-0.6 eV), in duralumin – 0.59 eV, and the bonding energy of solute atoms dissolved in aluminium with a vacancy equals: Mg to 0.46 - 0.6 eV, Si – 0.3-0.5 eV, Cu – 0.15-0.25 eV, Zn – 0.18 eV.



Figure 1: Hot shortness value distribution (mm) for the alloys of the aluminium angle sector of the Al-Cu-Mg system and phase areas [phase areas – 10-11] at the solidus temperature.

In reference [23] it is demonstrated that with the addition of 1.2-8.2% at. of Mg the loop density and vacancy concentration after quenching and ageing decreases by 100 times. Moreover, element atoms that have high bonding energy with vacancies also exert heavy influence on the kinetics [24]: for the alloy Al+0.5% at. Mg, aged at the room temperature, the time necessary for loop formation increases two-fold when compared with pure aluminium.

The Al-Cu alloys are noted for relatively rapid diffusion processes, since the Cu available has comparatively low vacancy bonding energy and is unable to influence significantly the kinetics of precipitation. These alloys age quickly and have a strong age-hardening response.

During solid solution decomposition, the θ phase is precipitated on the sub- and micrograin boundaries, which causes sizeable precipitate-free zones to be formed along the boundaries and brings about decrease in the alloy corrosion resistance.



Figure 2: Fluidity value distribution (mm) for the alloys of the aluminium angle sector of the Al-Cu-Mg system and phase areas [phase areas – 11] at the solidus temperature.



Figure 3: Polythermal sections of the aluminium angle sector of the Al-Cu-Mg system with the Cu content (% mass): a - 2.8, b - 4.2, c - 5.6

With an increase in alloying of elements Cu by Mg atoms, diffusion processes occur at a much reduced speed, solid solution decomposition slows down and evolution of disperse precipitation formation becomes difficult.

An increase in Mg content causes the S phase to appear and increase in volume fraction in the alloy microstructure. The nature of the S phase (Al₂CuMg) is based on the Cu and Mg content, and at the ratio of Cu/Mg (at.%) equal to 1 and less, there is only the S phase in the structure (in real commercial castings in some alloys there can be small quantity of the θ -phase due to non-equilibrium conditions of production).

As the Mg content increases, the potential for an ageing response drops drastically (for instance, alloys M40 and 1777 practically do not strengthen during exposure at 20°C, and

during heating they are not liable to respond) and the formation of precipitate-free zones along the micrograin boundaries is practically absent.

In Figure 4 a chart is presented of positioning in the Al-Cu-Mg system of the basic commercial aluminium alloys produced by the European Union, USA and other developed countries [12]. Table 1 contains the ratio value of Cu/Mg (at.%) for these alloys; at a ratio close to 1 and less, the S phase prevails in the structure and the alloys heat resistance is enhanced.

When examining the strength and heat resistance of alloys alloyed by Cu and Mg as presented on Figure 5 and 6, two basic facts can be pointed out: <u>firstly</u> – as the Mg content in the alloys increases (for example, at the constant Cu content of 3-4% mass), the strength and heat resistance values first increase, reaching maximum at the equiatomic content of Cu and Mg, and then decrease; <u>secondly</u> – decreased values of the characteristics involved are present with alloys a Cu content less than ~ 3% mass and more than ~ 6% mass.

In the <u>first case</u>, the change in properties is brought about by the phase content of alloys: at the low Mg contents there is only the θ -phase in the structure, or the S-phase appears increase. Further increase in the Mg content leads to its superfluity with regard to Cu (in at.%). As a result, following solid solution decomposition and the formation of dispersed S-phase particles, the solid solution remains saturated with Mg atoms. In this condition, the residual saturation of the solid solution approximates to a certain extent by its nature binary Al-Mg alloy (Al-Mg), the solid solution lattice intensity increases, its thermal stability is reduced and there is a reduction in alloy heat resistance. With still higher increase in the Mg content (Figure3) the T-phase (Al₆CuMg₄) may be formed and this causes further significant decrease in alloy strength and heat resistance.

In the <u>second case</u>: the decrease in strength and heat resistance of alloy with Cu < 3% mass takes place due to low general concentration of the alloying elements (alloys are positioned below the polytherm, Figure4), low fractions of the strengthening θ and S phases, and a very low residual saturation of the solid solution (it is because of this that, for instance, alloys of the 2618 series require large additions of Fe and Ni).

Alloy designation		1	2001	2004	2094	2014, 2014A, 2214	2095	2039	2195	2005	2098	2117
	at. %	2	7,14	5,0	3,96	3,7	3,52	3,43	3,33	2,92	2,89	2,83
Cu/Mg value	% mass	3	17,5	12,0	9,6	8,56	8,5	8,33	8,0	7,1	7,0	6,93

Table 1: Ratio of	f the Cu and	d Mg cont	tents in com	mercial	alloys [12] of the	Al-Cu-M	g systen	n.
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1	2017	2018	2036	2017A	x2096	2030	2007	2037	2024A , 2524, 2026	2224A	2009, 2424
2	2,77	2,5	2,37	2,36	2,07	1,78	1,47	1,33	1,25	1,22	1,2
3	6,67	6,06	5,78	5,71	5,05	4,33	3,6	3,27	3,04	2,96	2,92

1	2024,	2034	2224,	2218,	2031	2008	2048	2080	2038	2006	2618,	2091,
	2124		2324	2002							2618A	2010
2	1,19	1,15	1,1	1,09	1,05	0,96	0,9	0,82	0,76	0,65	0,61	0,59
3	2,9	2,8	2,73	2,67	2,56	2,4	2,2	2,0	1,82	1,58	1,48	1,43

In alloys with increased Cu content (>6% mass) the total saturation by the alloying elements (Cu+Mg) significantly exceeds the maximum total solid solubility in the solid solution, which is determined by the polytherm (Figure4). In the microstructure of these alloys, superfluous primary phases (θ , S) form and contribute to alloy embrittlement, and quite negatively affect the strength and heat resistance. Of great importance for the heat resistance of aluminium wrought alloys is also thermal stability of solid solution. Solid solution lattice tension occurring during alloying can exert significant influence on its thermostability. It is known that $r_{AI} = 0.143$ nm, $r_{Cu} = 0.128$ nm, and $r_{Mg} = 0.16$ nm, i.e. Cu atom radius is by 10.5% less, and Mg by 11.89% more than that of AI.



Figure 4: Layout of aluminium commercial wrought alloys in the Al-Cu-Mg system [12] and phase regions at the solidus temperature [10] (shaded is the area investigated in VIAM):

1* Alloys containing (% mass) Cu – 3.5...5.0; Mg – 0.2...1.0 (2014, 2014A, 2214, 2095, 2195, 2039, 2005, 2017, 2017A, 2018A);

2* - alloys containing (% mass) Cu – 3.2...4.5; Mg – 0.2...1.8 (2030, 2007, 2009, 2098); 3* - alloys containing (% mass) Cu – 3.5...5.0; Mg – 1.2...1.8 (2218, 2024, 2024A, 2124, 2224, 2224A, 2324, 2424, 2524, 2026, 2034).

As a result of alloying of aluminium with ~ 6%Cu (mass) the matrix lattice parameter is reduced to ~ 0.4038 nm (a_{Al} – 0.4049 nm), which brings about an increase in internal stress. With Mg addition to the Al-Cu alloys the lattice parameter value gradually increases and becomes practically equal to the pure aluminium when the Cu and Mg content is equiatomic. The internal tension of the lattice significantly decreased, and the alloy thermal stability and heat resistance both increase. Comprehensive investigations on alloys of the Al-Cu-Mg system (Figure1,2,4) carried out by the author has enabled development of a number of weldable alloys with enhanced heat resistance: 1150, 1151, 1153, and 1177. These alloys have no equal, they possess high heat resistance, good weldability and corrosion resistance. Detailed study of different properties of alloy 1151 [21, 25] showed significant advantage of both the base material and weldable joints (Figure 7) as compared with the existing (weldable heat-resistant) alloys.



Figure 5: Ultimate strength (MPa) at 20°C of as-naturally-aged aluminium alloys of the Al-Cu-Mg system as a function on of Cu and Mg content (% mass) [15,16].

Figure 6: Stress rupture values (h) of wrought alloys of the Al-Cu-Mg system as a function of Cu and Mg (% mass) content [17]: σ = 157 MPa; t = 200°C.

The strength of alloy 1151 weldable joints is no less than 80% of the base material strength (irrespective of a semi-finished product type and the welded seam direction). The hot cracking coefficient during welding of all the new alloys mentioned is no more than 5, and fluidity index amounts to 370-410.



Figure 7: Change in ultimate strength values of the base material and welded joint of 2219 and 1151 alloy sheets of 2-3mm at 20 - 450°C.

From the results of investigation into the structure and properties of the alloy 1151, the range of its application has been established: structures operating long-term (up to 50,000 h) at temperature up to 175°C, at 200°C for up to 100 h, and short-term at temperature up to 450°C. Increased Mg and S-phase content in duralumin alloys leads to the slowing of diffusion processes, increase in solid solution thermal stability, even distribution of fine-scale strengthening particles, and these factors enhance resistance of these alloys to creep and improve stress rupture properties.

The brief account of results of investigations into alloys of the Al-Cu-Mg system has demonstrated that there exist optimum contents of Cu and Mg which make it possible to

produce weldable aluminium alloys with enhanced heat resistance and that the most thermally stable and heat-resistant strengthening phase is the S-phase (Al₂CuMg).

References:

- [1] L. P. Luzhnikov. Deformiruemiye aluminieviye splavy dlya raboty pri povieshenniyekh temperaturakh, Metallurgia, Moskva, 290,1965.
- [2] O.A. Romanova. VINITI AN SSSR, Moskva, 30, 1958.
- [3] I.N. Fridlyander. Aluminieviye deformiruemiye konstruktsionniye splavy, Metallurgia, Moskva, 208, 1979.
- [4] F.I. Kvasov, I.N. Fridlyander. Aluminieviye splavy tipa duralumin, Metallurgia, Moskva, 240, 1984.
- [5] A.F. Lashko, C.V. Lashko-Avakian. Svarivaemii liyogkiye splavy, Sudpromgiz, Leningrad, 300, 1960.
- [6] I.I. Novikov. Goriyachelomkost tsvetnikh metallov i splavov. Nayka, Moskva, 298, 1966.
- [7] E.F. Chirkov. Izvestiy AN SSSR, Metalli, Moskva, 175, 1983.
- [8] E.F. Chirkov. Proceedings of 5-th Conf. Al. Alloys (ICAA-5), Part 1, Grenoble, France, 265, 1996.
- [9] E.F. Chirkov Aviatsionnie materiali i tekhnologii. Perspectivniye aluminieviye, magnieviye i titanoviye splavy dlya aviatsionnoy tekhiki, VIAM, Moskva, 104, 2001.
- [10] D.A. Petrov. Voprossi teorii splavov aluminiya, Metellurgizdat, Moskva, 256,1951.
- [11] H.W.L. Phillips. Inst. Metals, Monograph and Report Series N25, London, 86, 1959.
- [12] International Alloy Designations and Chemical Composition Limits for Wrought Aluminum and Wrought Aluminum Alloys. Unified North American and International Registration Records. Washington, 2001.
- [13] E.F. Chirkov. TLS, №4, Moskva, 64, 2002.
- [14] I.F. Kolobnev. Zharoprochnost liteynikh aluminievikh splavov, Metallurgia, Moskva, 320, 1973.
- [15] V.A. Livanov, E.I. Shilova i dr. Aluminieviye splavi. Promishlenniye deformiruemiye, spechioniye i liteyniye splavi. Metallurgia, Moskva,. 88, 1972.
- [16] M. Hansen und K.Z.Z Dreier. f. 14, Bd. 30, N6, 55, 1939.
- [17] E.I Shilova, O.G. Nikitayeva i dr. Metallovedeniye legkijkh splavov. Nayka, Moskva, 78, 1965.
- [18] L.F. Mondolfo. Aluminium Alloys: Structure and Properties. London-Boston, 407. 1976.
- [19] E.F. Chirkov. Metallovedeniye legkikh splavov. Nayka, Moskva, 63, 1970.
- [20] E.F. Chirkov. Aluminieviye splavi. Promishlenniye deformiruemiye, spechioniye i liteyniye splavi. Metallurgia, Moskva, 123, 1972.
- [21] Evgeny F. Chirkov, Iosif N. Fridlyander, Victor V. Cherkassov. Proceeding of 6-th Conf. Al. Alloys (ICAA-6), v3, Toyohashi, Japan, 2041, 1998.
- [22] R.N. Welson, I.E. Forsith. J.Inst. Met., 94, London, 8, 1966.
- [23] G. Thomas. Phis. Mag., 4, London, 1213, 1959.
- [24] J. Takamura, J.G.J. Greenfild. Appl. Phys. Mag., 33, 247, 1962.
- [25] E.F. Chirkov. Trudy pervoy mezhdunarodnoy konferentsii, Ros, Ingin. Akad., 5, Moskva, .27, 1995.