Influence of thermal treatment on the phase transitions of high-strength laser welding joints of aluminium-lithium alloys

A G Malikov¹, A M Orishich¹ and N V Bulina¹,²

¹ Khristianovich Institute of Theoretical and Applied Mechanics, Siberian Branch, Russian Academy of Sciences, 630090, Novosibirsk, Russia
² Institute of Solid State Chemistry and Mechanochemistry, Siberian Branch, Russian Academy of Sciences, 630128, Novosibirsk, Russia

E-mail: smalik@ngs.ru, laser@itam.ncs.ru

Abstract. The paper presents the results of analysis of the laser welding and following thermal treatment influence on the phase transitions and strength characteristics of aluminium-lithium alloy of the system Al-Mg-Li. The influence of the welded joints after-treatment (quenching and artificial ageing) on the structural and phase composition has been studied for different modes. It is shown that the quenching procedure changes the microstructure of the basic alloy and welded joint, and the mechanical characteristics in the alloy and welded joint were successfully equalized, so the sample composition was maximally homogenized.

1. Introduction

Development of aviation industry, both in Russia and abroad is inextricably connected with the task of increasing the productivity of aircraft assembly and reducing the environmental impact (reduction of CO₂ and other pollutants emission). There are two ways to solve this task in the aircraft fuselage assembly: reduction of weight (materials, rivets) and optimization of the aerodynamic structure. Weight reduction is possible owing to new aluminum-lithium alloys. The aluminum-lithium alloys have low density, increased elasticity module and high strength compared to ordinary aluminum alloys. In recent years, interest to medium-strength aluminum-lithium alloys of the system Al-Mg-Li has grown; these alloys are corrosion and crack-resistant, have extremely low density and are reliably weldable [1, 2]. Today, the urgent topic is to alternate riveting by welding in the aviation industry, thus the problem of development of modern welding methods for modern aluminum alloys possessing high mechanical characteristics of the welded connection is to be solved. New methods of connection of the parts from aluminum alloys are developed; they are based on melting welding (laser welding, argon-arc welding, electron-beam welding) and on friction stir welding. Laser welding has a number of advantages over the other methods of welding namely high process rate and small zone of thermal affect (hence, the decreased distortion degree). The major difficulty of the introduction of the laser welding of aluminum alloys is good mechanical properties of the welded connection approaching the ones of the basic alloy. But few works analyzing the welded connections of the alloys of the system Al-Mg-Li show that the strength of the welded joints is low and reaches only 0.67 – 0.8 of the basic alloy strength; the strength depends on the structural and phase transformations resulting from the high-speed heating, melting and following crystallization during the laser welding [3, 4].

As a continuation of [5,6], where we showed the promising future of the thermal treatment for the improvement of the mechanical properties, this paper analyzes the influence of quenching and
artificial ageing at different temperature and time modes on the phase transitions occurring in the alloy and welded joint. The paper presents the research of the thermal treatment on the microstructure and strength of the alloys’ welds 1424 (Al-Mg-Li system) developed in FSUE VIAM Russia and protected by Russian patents No. 2133295 [7,8]. The alloy 1424 is one of the most promising alloys for the welded fuselage of airbuses [9].

2. Experimental Technique
Laser welding was performed on the Sibir-1 ALTC [10]. The chemical composition (%, weight) of the alloy 1424 was Mg 5.4, Li 1.61, Zn 0.7, Zr 0.09, Sc 0.07 and Al balance [7,11]. The spectral analysis of the welded joint and base alloy was performed on a LEO 1430 VPI scanning electron microscope equipped with an IPX Oxford energy detector. Heat treatment was performed in a Carbolite batch furnace equipped with a temperature controller. The strength of the welded joints was measured at the static extension on the electro-mechanical test machine Zwick/Roell Z100. XRD patterns were recorded by a D8 Advance powder diffractometer. The XRD structure analysis was performed by the Rietveld method using Topas 4.2 software (Bruker AXS, Germany). Laser beam welding joints samples were obtained in the following welding mode: laser power 3 kW, welding speed 4m/min, focus forced into the material -3mm. After the welding the samples underwent heat treatment. The samples were tempered by heating to 450 °C, 490 °C at a rate of 5 °C/min with a 30-minute exposure followed by quenching in cold water. Artificial aging was carried out at temperatures of 120 °C and 175 °C, with aging time varying between 12 and 16 hours.

3. Results and Discussion
The maximal mechanical characteristics of the samples with the welded joint are reached after quenching at $T = 490$ °C and artificial ageing at 175 °C for 16 hours. Then, the microstructure is analyzed at these modes of the thermal treatment. The alloy 1424 is based on the $\alpha$-solid solution of alloying elements in aluminum ($\alpha$-Al), plus there are admixture phases $\delta'(Al_3Li)$, $S_1(Al_3MgLi)$ and coherent phases of intermetallics $Al_3Sc$ and $Al_3Zr$. In the microstructure seen from the electron microscope of the initial alloy, there are chains of agglomerates located predominantly on dendritic grain boundaries (Fig. 1a). On dendritic grains boundaries and inside them, many agglomerates form, their characteristic size of 0.5 – 1 µm.

![Figure 1](image_url)

Figure 1. SEM images of the alloy cross section (a, b, c) and joint cross section (d, e, f) without thermal treatment and with it. Magnification 1 000х Quenching (b, e), artificial ageing (c, f).
The photos made inside the dendritic grain of the initial alloy clearly show the particles of about 25 – 60 nm and 100 – 300 nm (Fig. 2a), which are distributed uniformly over the volume. The welded joint microstructure differs fundamentally (see Fig. 1d). In the solid solution of the welded joint, there are no particles of 25 – 60 nm, there are ones of 100 – 300 nm.

![SEM images](image)

**Figure 2.** SEM images of the alloy cross section (a, b, c) and joint cross section (d, e, f) without thermal treatment and with it. Magnification of 30 000х. Quenching (b, e), artificial ageing (c, f).

The thermal treatment changes the surface morphology. After the quenching, a number of particles decreased both in the alloy and joint (Fig. 1b, e). It is evident from Fig. 1 and 2 (c, f) that the ageing leads to the essential change in the basic alloy and joint microstructures. The microstructure of the thermally treated alloy approached the alloy in the delivery condition. Note that some changes related to the partial localization of agglomerated on dendritic grain boundaries take place in the joint microstructure after ageing, which has brought its structure closer to the structure of the alloy in the delivery condition (Fig. 1f). Blow up also demonstrates the formation of similar surface morphology in the initial alloy (Fig. 2c). Analyzing the welded joint and alloy microstructures, the following can be assumed. Agglomerate chains localized predominantly on the dendritic grains boundaries for the alloy (Fig. 1a) and both on the boundaries and inside the dendritic grains of the welded joint (Fig. 1d), are the assembly of the phase $S_1(Al_2MgLi)$. In the solid alloy solution inside the dendritic grain (Fig. 2a), the particles of about 25 – 60 nm are $\delta'(Al_3Li)$, whereas the ones of 100 – 300 nm belong to $S_1(Al_2MgLi)$. The triple phase $S_1(Al_2MgLi)$ is inside the welded joint dendrite, whereas the phase $\delta'(Al_3Li)$ is absent (Fig. 2d).

According to the XR analysis, the reflexes of these phases are registered in the alloy similar to diffractometry investigations on the reflection. The particles of the phase $\delta'(Al_3Li)$ are small, which causes the essential angular width of the diffraction reflex of this phase and, hence, the reduced intensity of the measured signal. The XR analysis was carried out both for the non-optimal mode of thermal treatment – quenching at $T = 450\,^\circ C$ (w2) and artificial ageing at 120 $^\circ C + 12$ hours (w3), the welded sample strength of 438 MPa [10], and in the optimal mode of the thermal treatment: $T = 490\,^\circ C$ (w4) + artificial ageing at 175 $^\circ C$ for 16 hours (w5), the strength being 500 MPa. The same is for the alloy before and after the optimal thermal treatment (a1-3, see Table 1).
Table 1. Structural characteristics of the solid alloy and welded joint solution.

|   | \(\alpha_1\)-Al |   | \(\alpha_2\)-Al |   | \(\alpha_3\)-Al |
|---|---|---|---|---|---|
| C, wt% | a, A | Cry Size, nm | C, wt% | a, A | Cry Size, nm | C, wt% | a, A | Cry Size, nm |
| a1 | - | - | 100 | 4.0705 | 121 | - | - | - |
| a2 | - | - | 100 | 4.0713 | 104 | - | - | - |
| a3 | - | - | 100 | 4.0711 | 128 | - | - | - |
| w1 | 6 | 4.083 | 50 | 82 | 4.0711 | 120 | 11 | 4.060 | 120 |
| w2 | 2 | 4.075 | 50 | 98 | 4.0705 | 166 | - | - | - |
| w3 | 7 | 4.075 | 50 | 93 | 4.0694 | 187 | - | - | - |
| w4 | 18 | 4.075 | 50 | 82 | 4.0697 | 214 | - | - | - |
| w5 | 17 | 4.075 | 50 | 83 | 4.0690 | 214 | - | - | - |

The XRD results show that the phase \(\alpha_1\)-Al with the cubic structure Fm\(\bar{3}\)m (see Table 1), the basic intermetallic phase \(\delta'(\text{Al}_3\text{Li})\) and triple phase \(S_1(\text{Al}_2\text{MgLi})\) (see Fig. 3) being inserted, is the base of the solid solution of the initial alloy.

Figure 3. XRD patterns of the base alloy and welded joint.
1) alloy 1424 2) welded joint 3) welded joint optimal heat treatment.

The welded joint contains 3 phases of solid solution \(\alpha_1\)-Al, \(\alpha_2\)-Al and \(\alpha_3\)-Al with different periods of grating crystallization and crystallite sizes (see Table 1 w1). This circumstance can be caused by the nonequilibrium crystallization with the emission of volatile alloying elements of the intermetallic triple phase \(S_1(\text{Al}_2\text{MgLi})\). As the temperature goes down, crystallization with crystals \(\alpha_j\)-Al begins, which is dictated by the residual composition of the liquid metal. In our case, the triple phase \(S_1(\text{Al}_2\text{MgLi})\) forms in the welded joint as a result of peritectic reaction, and the particles of this phase are located chaotically over the solid solution (see Fig. 1d). This conclusion agrees with the data of the solid solution grating parameters (see Table 1). Separation of the triple phase \(S_1(\text{Al}_2\text{MgLi})\) leads to the solid solution depletion of magnesium, which results in the grating period reduction. This is the circumstance which apparently caused the appearance of the phase \(\alpha_2\)-Al (see Table 1 w1) during the crystallization and melting in the welded joint zone, which is also confirmed by the joint microstructure parameters.
To reach the maximal strength of the thermally strengthened alloys, it is necessary, to produce a certain intermediate nonequilibrium structure which corresponds to the initial stages of decay of the over-saturated solid solution, using the ordered heating cycles.

After the quenching, the meta-stable phase $S_1(Al_2MgLi)$ is diluted in the alloy (see Fig. 1b), and the strengthening phase $\delta'(Al_3Li)$ is partially diluted inside the dendrite (see Fig. 2b). This factor causes the reduction of the alloy strength after the quenching from 512 MPa to 371 MPa. In the welded joint after the quenching, the phase $S_1(Al_2MgLi)$ dilution is also registered (see Fig. 1c). The strengthening phase $\delta'(Al_3Li)$ is almost absent there. The solid solution phase $\alpha_3$-Al is absent in the welded joint after the quenching, and the amount of the phase $\alpha_1$-Al rises essentially with the increased grating period (Table 3 w2). Apparently, the alloying components of the diluted intermetallic phases caused the increase of the grating period of the aluminum solid solution, the phase $\alpha_2$-Al, formed after the welding, transmitted into the phase $\alpha_1$-Al and increased its concentration, and also resulted in the phase $\alpha_7$-Al. The very important regularity is worth noting. The sample strength was almost independent of the quenching temperature. Taking into account that the strength characteristics of the welded joint without thermal treatment and of the samples after quenching are close to each other ($\sigma_b = 370 – 390$ MPa), we can conclude that the influence of the triple phase $S_1(Al_2MgLi)$ on the strength is weak, but the relative elongation rises. However, the quenching temperature not only changes the quantity of the triple phases $S_1(Al_2MgLi)$, but also influences the phase composition of the solid solution. At high quenching temperatures $T = 490÷530$ °C, the phase $\alpha_3$-Al with the big grating period rises essentially (see Table 2), which indicates the growth of free Mg, diluted in the solid solution. Thus, the quenching at the optimal temperature has led to the formation of the nonequilibrium state – the over-saturated solid solution of the alloying elements Mg and Li in aluminum. Both in the alloy and welded joint, the mechanical characteristics were successfully equalized, and we have found that at the optimal quenching temperature $T = 490÷530$ °C the composition of the whole sample is maximally homogeneous. The artificial ageing provides the separation of the main strengthening intermetallic phase $\delta'(Al_3Li)$ both in the joint and alloy, which is proven by both the XR (see Fig. 3) and high resolution microscopy data (Fig. 2c, f). The triple phase $S_1(Al_2MgLi)$ is localized on the dendritic grains boundaries, partially in the joint and almost fully in the alloy (see Fig. 1c, f). It is noted in [12, 13] that in the alloys with close molar concentrations of Mg and Li, which is actually the alloy 1424, the separation of the main strengthening phase $\delta'(Al_3Li)$, i.e. binding of Li, results in the increased concentration of “free” Mg in the solid solution and growth of the grating period. It is necessary to note that, according to [14], the major effect imposed by the magnesium participating in the phase $\delta'(Al_3Li)$ separation, is the reduction of lithium solvability in the solid solution. In our case, separation of the phase $\delta'(Al_3Li)$ at the optimal mode of the artificial ageing causes the increased tensile limit of the welded connection, which reaches the level of the initial alloy. Thus, the increased content of Mg in the solid solution of the welded joint after the quenching (the phase $\alpha_3$-Al), apparently promotes the more effective formation of the main strengthening phase $\delta'(Al_3Li)$ in the joint at the artificial ageing stage, which actually causes, after all, the strength of the permanent joint equal to the strength of the initial alloy.

**Conclusions**

For the first time ever, the closed cycle of phase transformations has been realized in the aluminum-lithium alloy Al-Mg-Li system on the basis of the complex approach including the optimal laser welding and following thermal treatment (quenching and artificial ageing). The pioneering mechanical characteristics comparable to the alloy in the deliver conditions ($\sigma_{UTS}= 500$ MPa, which is about 0.98 of the basic alloy strength) have been achieved for the welded connections after the thermal treatment. The use of optimal heat treatment through quenching with subsequent artificial aging enables the formation of the $\delta'(Al_3Li)$ strengthening phase in the weld joint. Thus, the pioneering equally strong welded joints were obtained for the aluminum alloy owning to the variation of the welded joint phase composition resulting from the thermal treatment.
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References
[1] Prasad N, Gokhale A, Wanhill R 2013 Aluminum–Lithium Alloys: Processing, Properties, and Applications (Butterworth-Heinemann).
[2] Dursun T and Soutis C 2014 Mater. Des. 56 862–71
[3] Kashaev N, Ventzke V and Çam G 2018 J. Manuf. Process. 36 571–600
[4] Xiao R and Zhang X 2014 J. Manuf. Process. 16 166–75
[5] Orishich A M, Malikov A G, Karpov E V, Pavlov N A and Mesenzova I S 2018 J. Appl. Mech. Tech. Phys. 59 561–68
[6] Annin B D, Fomin V M, Karpov E V, Malikov A G and Orishich A M 2017 J. Appl. Mech. Tech. Phys. 58 939–46
[7] Lukina E A, Alekseev A A, Khokhlatova L B and Oglodkov M S 2014 Met. Sci. Heat Treat. 55 466–71
[8] Khokhlatova L B, Kolobnev N I, Oglodkov M S, Lukina E A and Sbitneva S V. 2012 Met. Sci. Heat Treat. 54 285–89.
[9] Fridlyander I N, Khokhlatova L B, Kolobnev N I, Rendiks K and Tempus G 2002 Met. Sci. Heat Treat. 44 3–8
[10] Malikov A G and Orishich A M 2018 J. Phys. Conf. Ser. 1128 012053
[11] Sidhar H, Martinez N Y, Mishra R S and Silvanus J 2016 Mater. Des. 106 146–52
[12] Betsofen S Y, Antipov V V. and Knyazev M I 2016 Russ. Metall. 4 326–41
[13] Katsikis S, Noble B and Harris S J 2008 Mater. Sci. Eng. A 485 613–20
[14] Deschamps A, Sigli C, Mourey T, de Geuser F, Lefebvre W and Davo B 2012 Acta Mater. 60 1917–28