
PROMISING MATERIALS
AND TECHNOLOGIES

Al–Cu–Li and Al–Mg–Li Alloys: Phase Composition, Texture, and Anisotropy of Mechanical Properties (Review)

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Abstract—The results of studying the phase transformations, the texture formation, and the anisotropy of the mechanical properties in Al–Cu–Li and Al–Mg–Li alloys are generalized. A technique and equations are developed to calculate the amounts of the S_1 (Al_2MgLi), T_1 (Al_2CuLi), and δ' (Al_3Li) phases. The fraction of the δ' phase in Al–Cu–Li alloys is shown to be significantly higher than in Al–Mg–Li alloys. Therefore, the role of the T_1 phase in the hardening of Al–Cu–Li alloys is thought to be overestimated, especially in alloys with more than 1.5% Li. A new model is proposed to describe the hardening of Al–Cu–Li alloys upon aging, and the results obtained with this model agree well with the experimental data. A texture, which is analogous to that in aluminum alloys, is shown to form in sheets semiproducts made of Al–Cu–Li and Al–Mg–Li alloys. The more pronounced anisotropy of the properties of lithium-containing aluminum alloys is caused by a significant fraction of the ordered coherent δ' phase, the deformation mechanism in which differs radically from that in the solid solution.

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1. INTRODUCTION

Al–Li alloys are widely used in aerospace engineering due to a unique combination of a low density, a high strength, and very high elastic moduli as compared to other aluminum alloys. It should be noted that every 1 wt % Li (hereafter, wt % unless otherwise specified) decreases the alloy density by 3% and increases Young's modulus by 5% [1]. Along with the works on increasing the service properties of aluminum–lithium alloys, researchers at VIAM also develop the next generation of materials for aircraft construction, namely, layered glass-reinforced aluminum plastics based on the Al–Cu–Li system. These plastics can increase the elastic modulus by 8–10% and the density of these materials by 5–7% [2, 3].

The first aluminum–lithium alloy was patented by Czochralski in 1927 [4, 5]. Grade 2020 alloy (Al–Cu–Li system with 4.5 Cu and 1.1 Li) was developed in Alcoa in 1960, and grade 1420 aluminum alloy (Al–Mg–Li system) having the lowest density and was patented [6]. These alloys belong to the first generation of aluminum–lithium alloys. They were improved due to the oil crisis in 1973. As a result, second-generation alloys were developed; they had a high lithium content (more than 2%) and were represented by grade 2090 alloy (2.2 Li, 2.6 Cu, 0.12 Zr) and grade 8090 alloy (2.5 Li, 1.3 Cu, 1.0 Mg, 0.12 Zr). However, the alloys of that group had not found wide application

because of the fact that their mechanical properties, mainly fracture toughness, were only weakly higher than those of the aluminum alloys of the traditional alloying system Al(–Zn)–Mg–Cu. The main disadvantages were anisotropy of the mechanical properties and long-term low-temperature heating (LLH) embrittlement.

The third-generation aluminum alloys have a high Cu/Li ratio as compared to the second-generation alloys and are additionally alloyed with silver and zinc. These alloys include grade 2099 alloy (1.6 Li, 2.8 Cu, 0.7 Zn, 0.3 Mg, 0.3 Mn, 0.1 Zr) and its Russian analog, grade 1461 alloy. Weldalite 049 alloy (6.3 Cu, 1.3 Li, 0.4 Ag, 0.4 Mg, 0.14 Zr) and its Russian analog, grade 1469 alloy (4.3 Cu, 1.4 Li, 0.5 Ag, 0.2 Sc), exhibit extremely high strength and plastic characteristics. These alloys are promising for civilian and military aviation due to a balanced set of the mechanical properties, especially fracture toughness (which is very important for aviation engineering). The mechanical properties of these alloys are highly sensitive to the following microstructural parameters: the size, the morphology, the crystallographic orientation, and the local misorientation of matrix grains and intermetallic particles. The heterogeneity of the chemical and phase compositions, the texture, and the distribution of particles in the volume of semiproducts are also important.

The results of studying Al–Li alloys obtained in the 1970s–1980s were generalized in monograph [7], and

the results obtained in the recent years were generalized in fundamental monograph [1]. Nevertheless, there exist a number of problems to be refined; first of all, these are the problems related to a quantitative estimation of the contributions of intermetallic phases to the formation of texture and mechanical properties and their anisotropy in lithium-containing alloys. This review is devoted to these problems.

2. MORPHOLOGY AND CRYSTAL STRUCTURE OF INTERMETALLIC PHASES

The structure and the phase composition of Al–Cu–Li alloys were investigated in [8–13]. The main hardening phases in these alloys are represented by δ' (Al_3Li), T_1 (Al_2CuLi), and Θ' (Al_2Cu). δ' -phase particles have an $L1_2$ cubic structure ($a = 0.401$ nm) and the orientation relationship (OR) $(100)_{\delta'} \parallel (100)_{\text{Al}}$, $[100]_{\delta'} \parallel [100]_{\text{Al}}$ with an Al matrix. The particles of this phase have the following two morphological types: spherical particles ≈ 10 nm in diameter and banded particles 5×25 nm in size. Θ' -phase precipitates were found to nucleate on Guinier–Preston (GP) zones or Θ' -phase (Al_2Cu) particles. The ternary T_1 phase has the hexagonal lattice (space group $P6/mmm$, $a = 0.496$, $c = 0.935$ nm) with OR $(001)_{T_1} \parallel (111)_{\text{Al}}$, $[100]_{T_1} \parallel [110]_{\text{Al}}$. The metastable Θ' phase has the tetragonal lattice ($a = 0.404$, $c = 0.58$ nm) with OR $(001)_{\Theta'} \parallel (001)_{\text{Al}}$, $[100]_{\Theta'} \parallel [100]_{\text{Al}}$. Apart from these phases, the following phases can exist: phase T_2 (Al_6CuLi) with the cubic lattice ($a = 1.3914$ nm), phase S_1 (Al_2CuMg) with the orthorhombic lattice ($a = 0.401$, $b = 0.925$, $c = 0.715$ nm) in magnesium-containing alloys, phase β' (Al_3Zr) in zirconium-containing alloys. As the δ' phase, the last phase is present in the form of a metastable cubic modification with the $L1_2$ lattice ($a = 0.405$ nm) in most cases. The σ phase with the lattice ($a = 0.831$ nm) and OR $\{100\}_{\sigma} \parallel \{100\}_{\text{Al}}$, $\langle 100 \rangle_{\sigma} \parallel \langle 100 \rangle_{\text{Al}}$ was detected in the Al–Cu–Li system (3.7 Cu, 1.5 Li, 0.50 Zn, 0.37 Mg, 0.30 Mn, 0.14 Zr) along with phases T_1 , Θ' , and δ' [13].

The authors of [14, 15] used transmission electron microscopy to study the structures of the intermetallic phases that precipitate upon aging of Al–Cu–Li (3.2% Cu, 1.6 and 2.4% Li) alloys. Ingots were rolled to a thickness of 0.15 mm, and the sheets were heated to form a solid solution at 550°C for 30 min, cooled in iced water, and aged under various conditions. It was found that the δ' phase existed in the alloy with 2.4% Li after quenching and in the alloy with 1.6% Li after aging [14]. Aging of both alloys at 100°C is accompanied by homogeneous nucleation of GP-1 zones irrespective of the presence of the δ' phase. The δ' -phase particles having precipitated upon quenching dissolve upon aging and then again precipitate by nucleating on GP-1 zones. High-resolution transmission electron

microscopy was used to detect the following two types of lenticular δ' -phase particles decorating GP-1 zones: particles with in-phase and antiphase arrangement of atomic layers of the same kind.

The authors of [15] showed that the lattice parameters of the Θ' phase in the alloy with 1.6% Li are identical to those of a binary Al–Cu alloy ($a = 0.40$ nm, $c = 0.58$ nm). The alloy with 2.4% Li contains the Θ' phase in the form of two atypical structures, which are designated as T'_B of types I and II. The T'_B phase of type I is isostructural to the Θ phase, has a larger lattice parameter ($c \approx 0.64$ nm), and has a habit plane that is parallel to the $\{100\}_{\alpha}$ plane of the matrix. The T'_B phase of type II has lattice parameters $a = 0.41$ and $c = 0.61$ nm, and its habit plane makes an angle of approximately 20° with the $\{100\}_{\alpha}$ plane of the matrix at retained coherency with the matrix. The precipitates that are parallel to the $\{111\}$ plane in the alloy with 1.6% Li, which are usually identified as the T_1 phase, have a lattice parameter $c = 0.87$ – 0.94 nm, and the lattice parameter of the T_1 phase is $c = 0.94$ nm. The authors of [15] suppose that, under lithium deficiency, the crystal structure of the T_1 phase changes continuously from the hexagonal into the orthorhombic structure of the Ω phase, which has lattice parameters $a = 0.496$, $b = 0.859$, and $c = 0.848$ nm and OR $(001)_{\Omega} \parallel (111)_{\alpha}$, $[010]_{\Omega} \parallel [\bar{1}10]_{\alpha}$.

Al–Mg–Li alloys, which are represented by grade 1420, 1421, and 1423 alloys (3.5–6.2% Mg, 1.5–2.2% Li), have been developed and are mainly applied in Russia. Grade 1424 alloy contains $\approx 0.1\%$ Sc along with magnesium and lithium. Only one commercial alloy of this alloying system was developed in the United States (grade 5091 alloy with 3.95% Mg and 1.3% Li). The δ' phase is the main hardening phase in Al–Mg–Li alloys, which also contain the ternary S_1 phase (Al_2MgLi) with a complex cubic lattice ($a = 2.02$ nm with $Z = 104$ formula units). The studies [16] of the phase precipitation in an Al–5Mg–1.8Li alloy showed that the magnesium content in the δ' phase was approximately equal to the magnesium content in the alloy, and the lithium content in the δ' phase was 18.5 at %. When comparing the simulation and experimental results, the authors of [16] concluded that the main effect caused by magnesium participating in the δ' phase precipitation is a decrease in the lithium solubility in an α solid solution. The influence of magnesium on the particle–matrix interface energy and the diffusion mobility of lithium in aluminum was estimated to be insignificant.

3. MECHANISMS OF INTERMETALLIC HARDENING OF Al–Cu–Li ALLOYS

The effect of the intermetallic phases that precipitate during heat treatment on the mechanical proper-

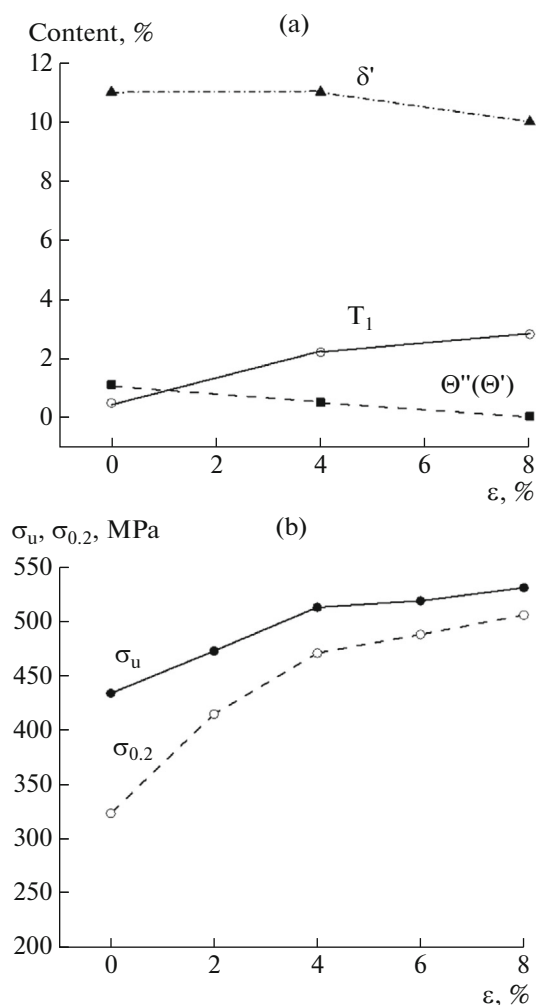


Fig. 1. (a) Contents of the intermetallic phases having precipitated upon aging (150°C, 24 h) and (b) strength of an Al–Cu–Li–X alloy vs. the strain during preliminary stretch flattening [19].

ties of Al–Cu–Li alloys was studied in [17–22]. The authors of [17, 18] used differential scanning calorimetry (DSC) and small-angle X-ray scattering to determine the phase contents. A quantitative correlation between the aging-induced hardening and the T_1 -phase particle sizes (plate diameter and thickness) was found in the AA2198 alloy (2.9–3.5 Cu, 0.8–1.1 Li, 0.25–0.8 Mg, 0.1–0.5 Ag, 0.04–0.18 Zr) aged at 155°C after preliminary quenching and deformation by 0.5–12% [17]. The hardening of the alloy was assumed to be only determined by the T_1 phase, which is generally accepted. Nevertheless, the existing experimental data allows another interpretation, which consists in the fact that the δ' phase mainly contributes to hardening upon artificial aging. In particular, GP zones, or the Θ and Θ' phases, are known to precipitate first among all copper-containing phases. Moreover, the balance of chemical and phase compositions implies the presence of the δ' phase (Al_3Li) in

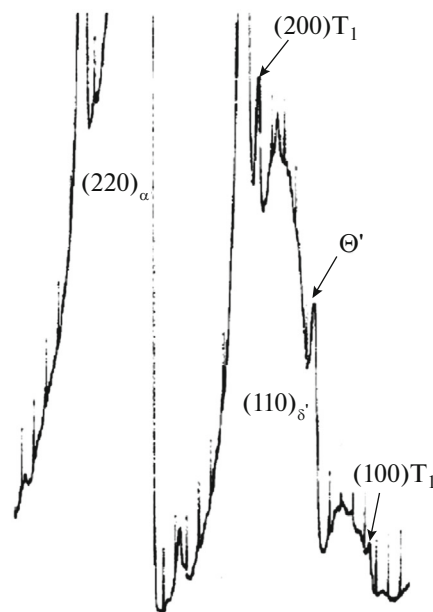


Fig. 2. Fragment of the X-ray diffraction pattern of the central layer in a 80-mm-thick plate made of a grade 1461 alloy after solid-solution treatment at 540°C for 1 h, tensile deformation by 2–3%, and steplike aging (120°C, 20 h + 140°C, 24 h + 150°C, 24 h).

the alloy. Indeed, the lithium content in the alloy is 3–4 at % (0.8–1.1 wt %) and the copper content is only 1.2–1.5 at % (2.9–3.5 wt %). Since the atomic fractions of lithium and copper in the T_1 phase (Al_2CuLi) are the same, remaining 1.5–2.8 at % Li cannot dissolve completely in a solid solution (equilibrium lithium concentration in the solid solution at room temperature is 1.5 at %) and lithium should precipitate in the form of the δ' phase. This precipitation is also promoted by the fact that part of copper should be involved in the solid solution or binary copper-containing phases, such as Θ and Θ' .

We now consider the problem of hardening phases in detail. The results of analysis of the influence of deformation during stretch flattening after solid-solution heat treatment on the amounts of the δ' , T_1 , and Θ' phases and the mechanical properties of an Al–Cu–Li–X alloy aged at 150°C for 24 h demonstrate that the strength characteristics of the alloy increase substantially with the tensile deformation, the amounts of the δ' and Θ' phases decrease, and the amount of the T_1 phase increases (Fig. 1) [19]. These results were unambiguously interpreted by Starke [19] in favor of the ternary T_1 phase as the dominating phase upon artificial aging despite the fact that its total volume fraction is lower than the δ' phase content by more than three times. According to our data, this difference is even larger.

Figure 2 shows part of the X-ray diffraction pattern (MoK α radiation) obtained by us for a sample cut from the median section of a 80-mm thick plate made

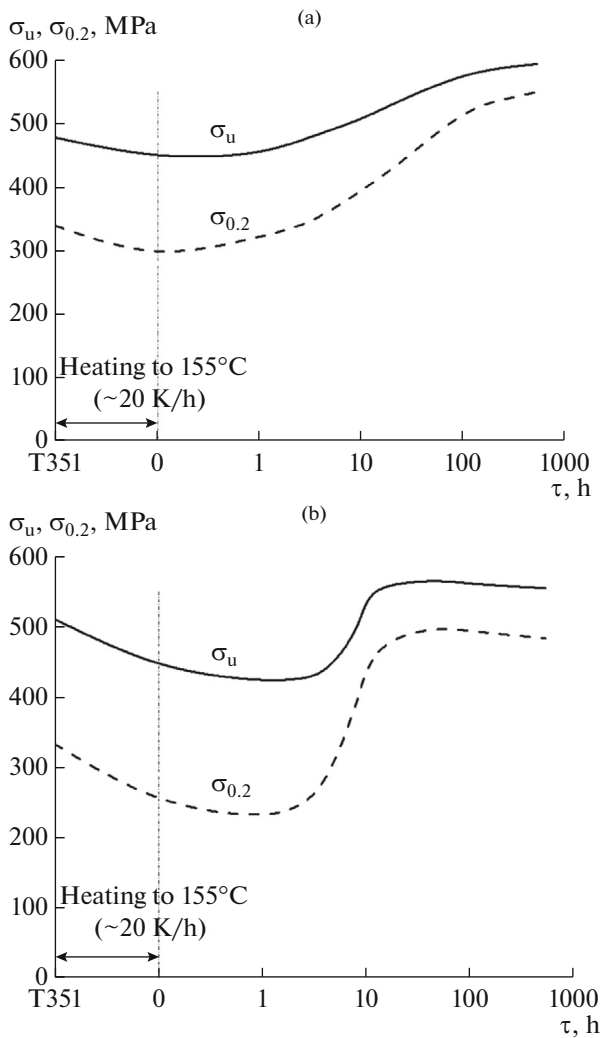


Fig. 3. Tensile strength of (a) AA2196 or (b) AA2198 alloy vs. the aging time. T351 means solid-solution treatment followed by natural aging [20].

of 1461 alloy (2.8 Cu, 1.7 Li, 0.5 Zn, 0.5 Mg, 0.1 Zr, 0.06 Sc) and subjected to solid-solution heat treatment at 540°C for 1 h, 2% tensile deformation, and three-step aging (120°C, 20 h + 140°C, 24 h + 160°C, 24 h). A quantitative relation between the phase contents can be estimated from the relative intensity of the (110) superlattice reflection of the δ' phase and the (100) and (200) reflections of the T_1 phase. According to the data in Fig. 2, the δ' phase is predominant. The plate in the median section has a strong single-component $\{011\}\langle 211 \rangle$ texture component in both the solid solution and the δ' phase, which explains the high intensity of its (110) superlattice reflection. However, texture $\{011\}\langle 211 \rangle$ is also preferred for the T_1 phase, since the (110) plane of the solid solution should be parallel to the (100) plane of the hexagonal phase according to the OR for the T_1 phase, in particular, $(001)_{T_1} \parallel (111)_{Al}$ and $[100]_{T_1} \parallel [110]_{Al}$. (A plane in the

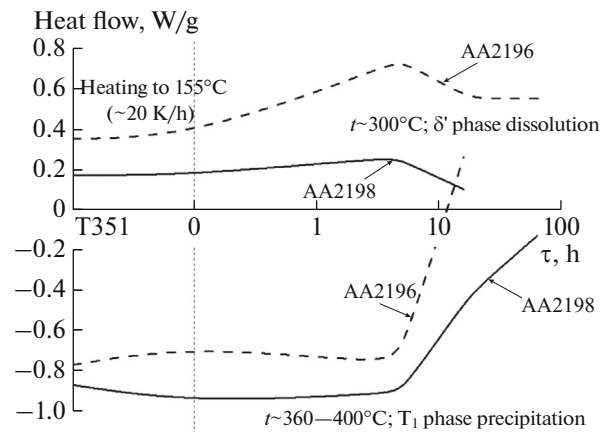


Fig. 4. DSC results for AA2196 and AA2198 alloys at various stages of aging. The peaks correspond to the endothermic dissolution of the δ' phase at $\approx 300^\circ\text{C}$ and the exothermic precipitation of the T_1 phase at 360–400°C [20].

cubic lattice is always normal to the corresponding direction, and a plane in the hexagonal lattice is normal to the corresponding direction only for the basal plane and the planes that are normal to it, including $\{100\}$.) The given OR is fulfilled for the T_1 phase, since even traces of the reflections of the T_1 phase were not detected for the sections of the plate where the $\{011\}\langle 211 \rangle$ texture component was absent.

Thus, a critical analysis of the results of experimental investigations suggests that the role of the T_1 phase as the main hardening component of Al–Cu–Li alloys was overestimated, especially for the alloys containing more than 1.5% Li. This conclusion also follows from [20], where the mechanical properties and the phase transformations of grade 2196 and 2198 alloys (Al–Cu–Li system) containing 2.9 and 3.2% Cu and 1.7 and 1% Li, respectively, were studied. Aging at 155°C after quenching and 2% tensile deformation of the grade 2196 alloy, where the δ' phase mainly precipitates, causes higher strength properties as compared to the grade 2198 alloy (Fig. 3a), where aging is mainly accompanied by the precipitation of the T_1 phase (Fig. 3b).

An analysis of the DSC curves of the alloys after various aging regimes showed that the T_1 phase precipitated in both alloys upon aging for about 5 h (Fig. 4). This is indicated by a sharp decrease in the precipitation of this phase in a DSC experiment after the corresponding aging time. Nevertheless, the hardening of the grade 2196 alloy occurs both before and after the precipitation of this phase, and this precipitation can only be related to the δ' phase (see Fig. 3a). The estimation performed by DSC and small-angle X-ray scattering showed that the maximum amount of the δ' phase ($\approx 10\%$) corresponded to the state after solid-solution treatment and natural aging (state T351). Slow heating to an aging temperature of

155°C leads to the dissolution of the fine δ' -phase particles having precipitated upon natural aging, and their fraction decreases from 10 to 4%. This decrease is accompanied by a noticeable decrease in the strength properties (see Fig. 3). The amount of the δ' phase then increases to $\approx 7\%$ in the first 5 h of aging and then levels off, and the strength properties grow continuously up to 1000 h of aging. The strength of the alloy grows most intensely upon aging for 10–100 h, when neither δ' nor T_1 precipitate (see Fig. 3a).

The hardening of the alloy in the absence of precipitation of intermetallic particles can only be explained by an increase in the δ' -phase particle sizes (Fig. 5), which is natural for any aging processes (the amount of an intermetallic phase can increase or decrease with time, and the particle size can only increase). This behavior was repeatedly observed in many works. For example, Setyukov [22] analyzed the broadening of X-ray diffraction lines and detected an increase in the δ' -phase particle size from 6 to 25 nm in 24 h aging at 150°C. The hardening mechanism caused by the coarsening of the δ' phase at a constant volume fraction of its particles has not been noted, since it is inherent only in the particles the hardening mechanism of which is only caused by ordering.

The hardening of binary Al–Li alloys is known to be related to the precipitation of fine particles of the metastable coherent ordered δ' phase (Al_3Li) with the $L1_2$ structure. These particles are cut by paired dislocations, so that the leading dislocation forms an anti-phase boundary (APB) and the next paired dislocation restores the order. The main hardening effect is associated with ordering, since the hardening effect of the δ' phase induced by an increase in the particle–matrix interface surface and the increase in the stresses due to the coherent boundary is negligibly small because of a low boundary energy (0.01–0.02 J/m²) and the lattice misfit strain is lower than 10^{-3} [23]. Coherent ordered particles harden the matrix by $\Delta\tau$,

$$\Delta\tau = \frac{\gamma}{2b} \left[\left(\frac{3\pi^2 \gamma f d}{64T} \right)^{1/2} - f \right], \quad (1)$$

where γ is the APB formation energy for the (111) slip plane, b is the Burgers vector, d is the particle diameter, f is the volume fraction of particles, and T is the linear dislocation tension.

With allowance for the Sachs factor ($M = \sigma/\tau = 2.24$), the hardening of a polycrystal is

$$\Delta\sigma = M\Delta\tau = 2.24\Delta\tau. \quad (2)$$

The linear tension is

$$T = \frac{Gb^2}{4\pi} \left(\frac{1 + \nu + 3\nu \sin^2 \xi}{1 - \nu} \right) \ln \left(\frac{L_s}{r_0} \right),$$

where ξ is the angle between the Burgers vector and the line of the initial rectilinear dislocation; L_s and r_0

are the outside and inside radii of curvature, respectively; and G is the shear modulus of the matrix.

Screw dislocations, which are predominant in Al–Li alloys, are characterized by $\xi = 0$ and $\nu = 1/3$; as a result, we have

$$T = \frac{Gb^2}{2\pi} \ln \left(\frac{L_s}{r_0} \right).$$

Parameter L_s is taken to be the interparticle distance and r_0 is taken to be b .

Equation (1) is valid for comparatively small particles, $d < 8T/\gamma\pi$. If particles are larger, the mechanism of interaction with dislocations changes and the critical reduced shear stress (CRSS) ceases to depend on the particle size and remains constant. The limiting particle size for which Eq. (1) holds true is $d = 40$ nm (400 Å). The Orowan mechanism begins to operate for larger particles, the limiting size of which is $d = 50$ nm.

The authors of [24] used this method of determining the hardening of binary Al–Li alloy induced by the precipitation of δ' -phase particles to estimate the hardening of ternary Al–Mg–Li alloys (this method can also be applied to Al–Cu–Li alloys). Substituting $G = 30$ GPa, $b = 0.286$ nm, $\gamma = 165$ mJ/m², and $T = 2.0$ nN into Eqs. (1) and (2), they estimated the hardening as a function of the size and the fraction of δ' -phase precipitates. These data are shown in Fig. 6. The maximum hardening (≈ 200 MPa) is seen to correspond to the limiting δ' -phase particle size of 40 nm and the fraction of particles of 12%. This estimate is close to the experimental hardening detected upon hardening of the grade 2196 alloy (see Fig. 3a).

It should be noted that the hardening induced by the precipitation of ordered phases depends on the APB formation energy, which is very sensitive to alloying. This finding can explain the substantial changes in the mechanical properties of Al–Cu–Li alloys that are caused by alloying with various elements, such as silver, zinc, and magnesium. It is very important that this effect is achieved upon alloying by the amount of an element that is too low for this element to influence the properties as a result of the formation of an individual phase.

The proposed mechanism of hardening aluminum–lithium alloys upon aging explains the necessity of their stretch flattening and low-temperature aging (120°C). These technological operations provide the optimum structure–phase state to achieve the maximum hardening at the next stage of aging at 140–160°C. In particular, cold deformation during tension favors the formation of δ' -phase and, apparently, Θ' -phase particles, which are resistant to the subsequent low-temperature aging. This aging causes the dissolution of the fine δ' -phase particles that have formed during natural aging and, according to Eq. (1), weakly contribute to the hardening. Here, the fraction of intermetallic phases decreases and the strength

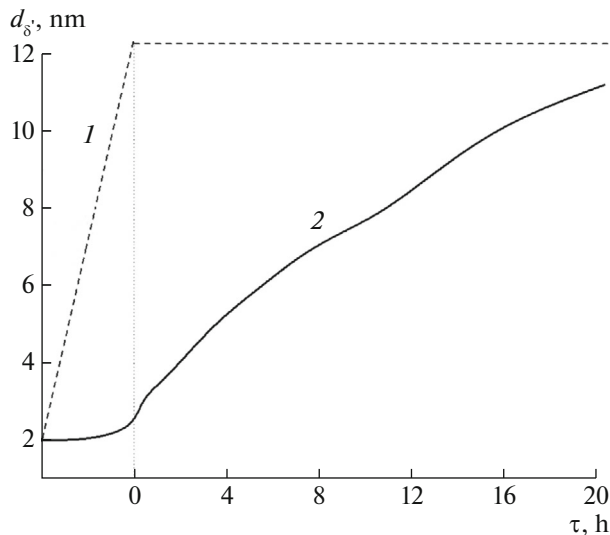


Fig. 5. Growth kinetics of δ' -phase particles during (1) slow (≈ 20 K/h) heating to 155°C and (2) aging at 155°C of an AA2196 alloy [20].

properties degrade (see Fig. 3). However, the number of δ' -phase nuclei that forms upon cold deformation and low-temperature aging is optimal to achieve the maximum hardening effect due to an increase in the particle size at the final stage of aging. Low-temperature aging can be effectively replaced by slow heating, which simplifies the technology (see Fig. 5). If the T_1 phase is assumed to be the main hardening phase during artificial aging, it is difficult to explain the thermomechanical treatment of Al–Cu–Li alloys used in real practice.

4. QUANTITATIVE PHASE ANALYSIS OF Al–Mg–Li AND Al–Cu–Li ALLOYS

One of the most important requirements imposed on aviation engineering materials is the stability of their mechanical properties over the entire period of operation (20 years) at slightly elevated temperatures (50 – 100°C). This requirement cannot be met by Al–Li alloys containing more than 1.7% Li (especially if an alloy is in an incompletely aged state). For example, the flow stress of binary alloys containing 2.5% Li increases by 10% and their impact toughness decrease by 43% upon aging at 100°C for 1000 h [25]. Similar results were obtained for ternary Al–Cu–Li and Al–Mg–Li and quaternary Al–Cu–Mg–Li alloys.

The character of changing the mechanical properties of the alloys depends on an alloying system. The authors of [25] studied the effect of magnesium and copper alloying of a binary Al–1.7Li alloy on the mechanical properties and phase stability of the alloys subjected to holding at 70°C for 1000 h. It was found that the holding-induced hardening of ternary Al–

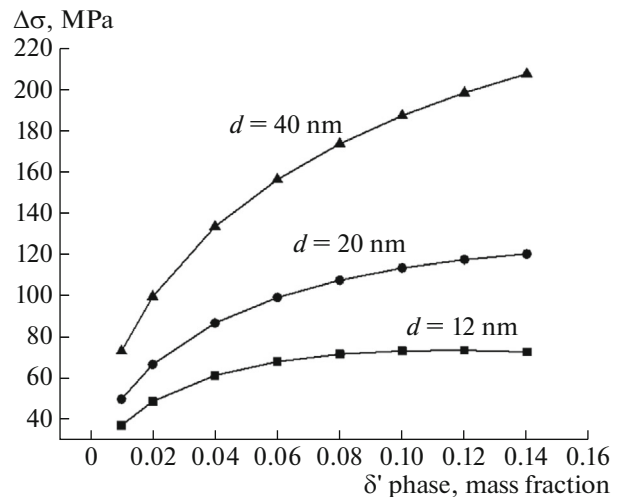


Fig. 6. Hardening of Al–Li alloys vs. the size and fraction of δ' -phase particles. Calculation by Eq. (2) [24].

1.7Li– x Cu alloys increased substantially with the copper content, and a uniform hardening of quaternary Al–1.2Mg–1.7Li– x Cu alloys was detected in the copper concentration range 1.2–3%.

The results of a transmission electron microscopy (TEM) investigation of binary Al–1.7Li alloys demonstrate that their aging was accompanied by the precipitation of fine δ' -phase particles. The formation of a mixture of the δ' phase and GP zones was detected in ternary Al–1.7Li–Cu alloys under similar conditions, and the formation of a mixture of the δ' phase and Cu/Mg clusters was observed in quaternary Al–1.2Mg–1.7Li–Cu alloys. The precipitation of the δ' phase was significantly enhanced due to its nucleation on GP zones and Cu/Mg clusters in the ternary and quaternary alloys, which led to a more significant hardening of these alloys as compared to the binary Al–Li alloys.

To calculate the anisotropy of the strength and elastic properties and to predict the thermal stability of Al–Li alloys, one has to have information on a quantitative relation between the intermetallic phases in them. Basic equations for calculating the changes in the volume and linear sizes of aluminum alloys that are induced by the dissolution (precipitation) of particles during heat treatment were presented in [26]. A technique for quantitative phase analysis of Al–Mg, Al–Cu, and Al–Mg–Li alloys, which is based on measuring the lattice parameter of a solid solution, was developed in [27] and was then extended to Al–Cu–Li alloys in [28, 29].

Two intermetallic phases, namely, δ' (Al_3Li) and S_1 (Al_2MgLi), are mainly present in grade 1420 alloys (Al–Mg–Li). Their quantitative ratio can be calcu-

lated from the lattice parameter of a solid solution [29]. Magnesium substantially increases the lattice parameter; therefore, the precipitation of the S_1 phase is accompanied by a decrease in this parameter. The lithium content in the solid solution weakly affects its lattice parameter; however, the precipitation of the δ' phase leads to magnesium enrichment of the solid solution and, hence, to an increase in its lattice parameter.

Under certain assumptions, Al–Cu–Li alloys can be considered as three-phase mixtures of an α solid solution and phases δ' and T_1 . Apart from these phases, these alloys can also contain copper-containing binary (Θ , Θ' , Θ'') and ternary (T_2 (Al_6CuLi), S' (Al_2CuMg)) phases and the β' (Al_3Zr) phase (in zirconium-containing alloys). However, their contributions are insignificant in quantitative calculations despite the fact that their role can be substantial, as in the case of precipitation of fine δ' -phase particles at GP-zone boundaries during LLH at 70°C [25]. We now write the following equations for the balance of the elemental and phase compositions of Al–Cu–Li alloys:

$$\begin{aligned} 100X_{Al}^0 &= X_{Al}^\alpha W_\alpha + X_{Al}^{T_1} W_{T_1} + X_{Al}^{\delta'} W_{\delta'}; \\ 100X_{Cu}^0 &= X_{Cu}^\alpha W_\alpha + X_{Cu}^{T_1} W_{T_1}; \\ 100X_{Li}^0 &= X_{Li}^\alpha W_\alpha + X_{Li}^{T_1} W_{T_1} + X_{Li}^{\delta'} W_{\delta'}; \\ W_\alpha + W_{T_1} + W_{\delta'} &= 100. \end{aligned} \quad (4)$$

Here, X_{Al}^0 , X_{Cu}^0 , and X_{Li}^0 are the aluminum, copper, and lithium contents (mass fractions) in an alloy, respectively; W_α , W_{T_1} , and $W_{\delta'}$ are the α , T_1 , and δ' phase contents (wt %), respectively; X_{Al}^α , X_{Cu}^α , X_{Li}^α , $X_{Al}^{T_1}$, $X_{Cu}^{T_1}$, $X_{Li}^{T_1}$, $X_{Al}^{\delta'}$, and $X_{Li}^{\delta'}$ are the aluminum, copper, and lithium contents (mass fractions) in the α , T_1 , and δ' phases, respectively.

The values of $X_{Al}^{T_1}$, $X_{Li}^{T_1}$, $X_{Al}^{\delta'}$, and $X_{Li}^{\delta'}$ are calculated with allowance for the stoichiometries of the T_1 (Al_2CuLi) and δ' (Al_3Li) phases. X_{Cu}^α is determined from the lattice parameter of the solid solution a_α according to Vegard's law,

$$X_{Cu}^\alpha = (a_\alpha - a_{Al}) / \frac{\Delta a_{Cu}^\alpha}{\Delta X_{Cu}^\alpha},$$

where $\frac{\Delta a_{Cu}^\alpha}{\Delta X_{Cu}^\alpha}$ is the change in the lattice parameter per 1 wt % alloying element, Å/% (the values are given in monograph [30]).

The solution to set (4), where the lithium content in the α phase (X_{Li}^α) is a variable parameter, can be written in the form

$$\begin{aligned} W_\alpha &= [(X_{Li}^{\delta'} - X_{Li}^{T_1})(X_{Al}^0 X_{Cu}^{T_1} - X_{Al}^{T_1} X_{Cu}^0) \\ &\quad - X_{Al}^{\delta'} X_{Cu}^{T_1} (X_{Li}^0 - X_{Li}^{T_1})] / [(X_{Li}^{\delta'} - X_{Li}^{T_1}) \\ &\quad \times (100X_{Cu}^{T_1} - X_{Cu}^\alpha X_{Cu}^{T_1} - X_{Cu}^{T_1} X_{Li}^\alpha - X_{Al}^{T_1} X_{Cu}^\alpha) \\ &\quad - X_{Al}^{\delta'} X_{Cu}^{T_1} (X_{Li}^\alpha - X_{Li}^{T_1})] \times 100, \end{aligned} \quad (5)$$

$$W_{T_1} = \frac{100X_{Cu}^0 - X_{Cu}^\alpha W_\alpha}{X_{Cu}^{T_1}}, \quad W_{\delta'} = 100 - W_\alpha - W_{T_1}.$$

The set of equations (5) can be used to determine the ratio of the α , T_1 , and δ' phases for any lithium content in the α phase.

For Al–Mg–Li alloys, the component contents in the S_1 phase ($X_{Al}^{S_1}$, $X_{Li}^{S_1}$, $X_{Mg}^{S_1}$) should be included in the set of equations (5) instead of the component contents in the ternary T_1 phase ($X_{Al}^{T_1}$, $X_{Li}^{T_1}$, $X_{Cu}^{T_1}$). X_{Mg}^α is determined according to Vegard's law similarly to X_{Cu}^α ,

$$X_{Mg}^\alpha = (a_\alpha - a_{Al}) / \left(\frac{\Delta a}{\Delta X} \right)_{Mg}^\alpha.$$

The calculation procedure was applied to determine the intermetallic-phase contents in ternary alloys containing copper or magnesium, in particular, Weldalite 049 (6.3 Cu, 1.3 Li) and 8090 (1.1 Cu, 2.4 Li) alloys and grade 1420 alloy (5.2 Mg, 1.6 Li). The calculation results are shown in Figs. 7 and 8.

The ratio of the phase contents in alloys is determined by the ratio of the lithium concentration to the copper (or magnesium) concentration. We now consider the copper-containing grade 8090 alloy. In the calculations, the lithium concentration in the solid solution of this alloy was taken to be zero ($X_{Li}^\alpha = 0$), which corresponds to the maximum amount of the δ' phase. It is clear that the maximum amount of the T_1 phase corresponds to the absence of copper in the solid solution and is determined by the molar copper concentration in the alloy and that the maximum amount of the δ' phase corresponds to the absence of lithium in the solid solution and is determined by the difference between the molar concentrations of lithium and copper, since the ternary phase has the composition Al_2CuLi . (If a certain amount of the Θ , Θ' , or Θ'' phase formed under these conditions, this amount would exert a little effect—only the amount of the T_1 phase would decrease.) Since the entire copper is present in the ternary phase (where every copper atom is related to a lithium atom), the difference between the molar concentrations of lithium and copper controls the maximum amount of the δ' phase in the alloy. Therefore, the maximum amounts of the T_1 and δ' phases in the Weldalite 049 alloy, which contains 4.8 at % Li and 2.7 at % Cu, are 12 and 18%, respec-

tively (see Fig. 7a). The maximum amounts of the T_1 and δ' phases in the grade 8090 alloy, which contains 8.7 at % Li and 0.4 at % Cu, are 2 and 30%, respectively (see Fig. 7b).

The curves in Figs. 7 and 8 for the δ' phase and the ternary phases with a similar stoichiometry, namely, T_1 (Al_2CuLi) in Al–Cu–Li alloys and S_1 (Al_2MgLi) in Al–Mg–Li alloys, are explained by the difference effects of copper and magnesium on the lattice parameter (copper decreases it and magnesium increases it). Therefore, the precipitation of the δ' phase leads to an increase in the lattice parameter of the α solid solution in the magnesium-containing alloys and to a decrease in this parameter on the copper-containing alloys due to the “extraction” of aluminum from the solid solution and, correspondingly, an increase in the magnesium or copper content in it. The precipitation of the ternary phase (S_1) in the magnesium-containing alloys results in a decrease in the lattice parameter of the solid solution, and that in the copper-containing alloys (T_1) results in its increase due to a decrease in the magnesium or copper content in it, respectively.

The ratio of the fractions of the δ' and T_1 (or S_1) phases is determined by the ratio of the molar concentrations Y of lithium and copper (or magnesium). This ratio in Al–Cu–Li alloys is $Y_{Li}^\alpha/Y_{Cu}^\alpha = 2-19$; therefore, the δ' phase is predominant in them (see Fig. 7). This ratio in Al–Mg–Li alloys is $Y_{Li}^\alpha/Y_{Mg}^\alpha = 1.0-1.5$; therefore, the amounts of the δ' and S_1 phases are approximately the same (see Fig. 8). The generalization of the results obtained for a number of well-known alloys gave an almost linear dependence of the relative fraction of the δ' or T_1 (S_1) phase on the ratio of the molar lithium and copper (magnesium) concentrations (Fig. 9).

5. TEXTURE AND ANISOTROPY OF THE PROPERTIES OF Al–Li ALLOYS

Most investigations of the texture of aluminum alloys have long been motivated by the requirements of the automotive industry to ensure high technological characteristics upon deep drawing of sheets [31–35]. Therefore, these investigations were focused on analyzing the effect of texture on the anisotropy characteristics, such as the Lankford coefficient and yield contours [36]. Interest in texture studies of aluminum alloys in the aerospace industry is mainly caused by the anisotropy of the elastic moduli and the strength properties.

The texture of cold rolling of aluminum sheets is described by the following components, which are typical of fcc metals: copper texture $\{112\}\langle 111 \rangle$, brass texture $\{110\}\langle 112 \rangle$, and S texture. The last texture includes the orientations for which the normals to the rolling plane (ND) are located between the $\{110\}$ and $\{112\}$ plane poles in the stereographic triangle, e.g.,

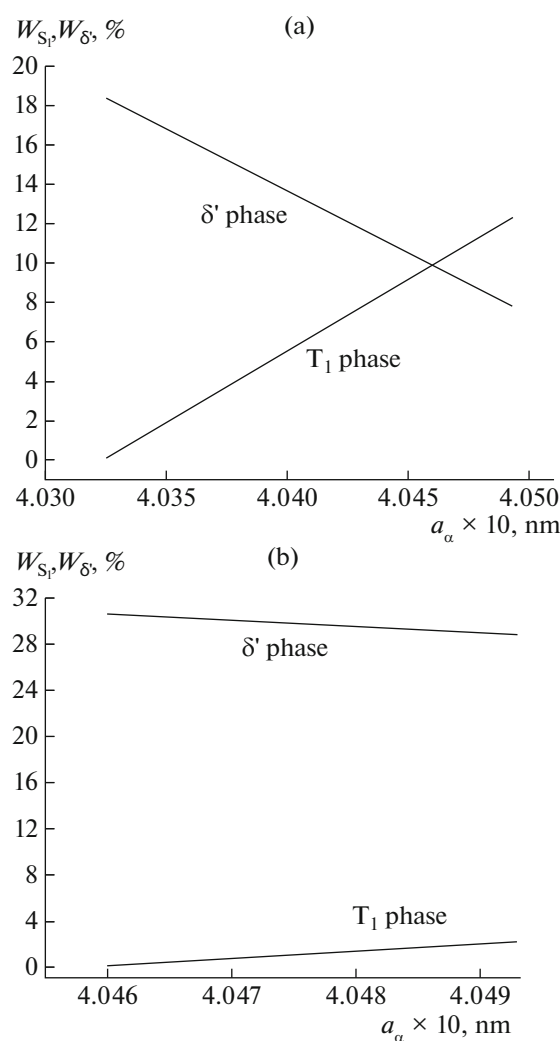


Fig. 7. Effect of the ratio of δ' to T_1 in alloys (a) Weldalite 049 and (b) grade 8090 on the lattice parameter of the solid solution at $X_{Li}^\alpha = 0$ [29].

$\{123\}\langle 634 \rangle$. The minor rolling texture components in aluminum include Goss texture $\{110\}\langle 100 \rangle$. Cube texture $\{001\}\langle 100 \rangle$ is the predominant recrystallization texture of aluminum, and another well-known recrystallization texture is R texture. The last texture is close to rolling texture S and it is sometimes identified as the S component partly retained upon recrystallization. The surface layers contain shear textures $\{001\}\langle 110 \rangle$ and $\{111\}\langle 110 \rangle$.

The textures of aluminum–lithium alloys do not differ radically from those of aluminum alloys. However, some authors relate a higher anisotropy of the mechanical properties of lithium-containing aluminum alloys to the pronounced single-component $\{110\}\langle 112 \rangle$ brass texture. The formation of this component was attributed to the suppression of cross sliding due to the splitting of dislocations in copper–zinc

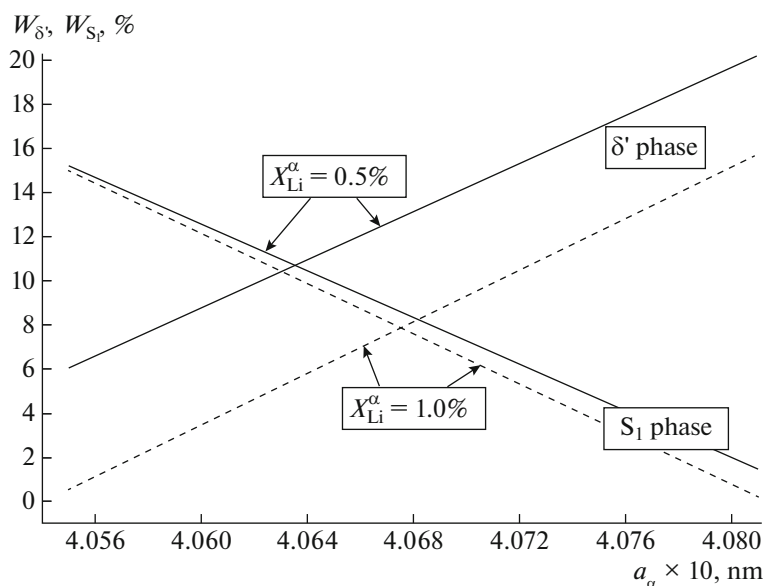


Fig. 8. Effect of the contents of the δ' and S_1 phases in grade 1420 alloy on the lattice parameter of the solid solution at a lithium content X_{Li}^α = (solid line) 0.5 and (dashed line) 1 wt %.

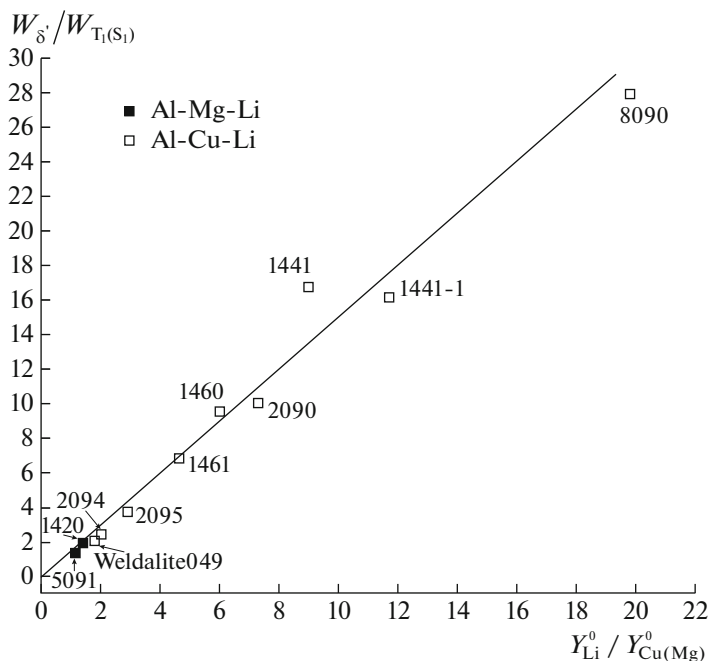


Fig. 9. Ratio of the mass fractions of the δ' , S_1 , and T_1 phases and the molar concentrations of lithium, copper, and magnesium in Al–Cu–Li and Al–Mg–Li alloys [29].

alloys (brass) with a low stacking-fault energy (SFE). As follows from this standpoint, the copper texture should dominate in aluminum with a high SFE. However, experimental studies of the textures in sheet semiproducts do not support this assumption: in most cases, the brass texture predominates over the copper texture, as in the case of other fcc metals.

The influence of temperature and strain on the formation of hot rolling texture in two typical aerospace alloys, namely, grade AA2050 (Al–Cu–Li system) and grade AA7050 (Al–Zn–Mg–Cu) alloys, was comprehensively studied in [37]. A strong brass texture was found to form in both alloys; that is, this texture is independent of the presence of lithium. As the defor-

mation temperature increases, brass texture component $\{110\}\langle 112 \rangle$ grows, which is associated with increasing sliding activity on the systems that differ from $\{111\}$.

When analyzing the diffraction patterns of individual grains in a strongly textured grade AA2195 alloy, the authors of [38] revealed differences in the states of stress in grains of different orientations. These differences are likely to be related to the orientation dependence of the phase transformations in these grains and, correspondingly, the differences in the solid-solution compositions and the fractions of precipitating phases in them. The texture of hot rolling of ternary Al–Li–Cu alloys as a function of the ratio of the mass fractions of lithium and copper was studied in [39].

The results [40] of studying the influence of a change in the rolling direction (cross rolling) on the formation of rolling and annealing texture in an Al–Zn–Mg–Cu–Zr alloy are interesting. It was found that “rotated” brass texture $\{110\}\langle 556 \rangle$, which is close to fourfold symmetry (which is the cross rolling condition), formed after rolling. The development of this texture component depended on recrystallization occurring in intermediate annealing. In particular, an electron backscatter diffraction investigation showed that recrystallized grains after rolling had the cube orientation or the rotated cube orientation and that fine Al_3Zr particles stabilized the orientations that were close to the brass texture.

The brass texture is most intense in the alloys that contain the particles hindering grain growth. The non-uniformity of deformation in grains with different orientations is likely to play an important role, with particles precipitating most intensely in grains with the maximum effective deformation. This deformation increases in the row $\langle 100 \rangle + \langle 111 \rangle \rightarrow \langle 112 \rangle \rightarrow \langle 135 \rangle \rightarrow \langle 110 \rangle$; that is, it is maximal in the grains the orientations of which are close to the brass texture. Recrystallization nuclei form in these grains first of all; however, their high density hinders grain growth. The grains that have a lower effective deformation and, hence, a lower particle density grow more rapidly.

If the total number of particles is small, coarse recrystallized grains with a cube orientation absorb all deformed grains, including those with the brass texture, during growth. In this case, the cube texture, which is typical of the recrystallization texture of pure aluminum, forms. If the number of particles is large, the growth of recrystallized grains is limited and they cannot absorb the grains of the main texture component. As a result, the brass texture remains predominant, but its intensity can weaken because of partial absorption by recrystallized grains. As a result, a compromise recrystallization texture forms by analogy with the deep drawing texture in steel [41]: it is known that the recrystallization texture in an automobile-body sheet is also controlled by interstitial-phase particles (mainly niobium and titanium carbonitrides)

rather than intermetallic phases, as in aluminum alloys.

The contribution of texture to the anisotropy of the mechanical properties depends on the technological history of the material. The results of estimating this contribution using the Taylor factors calculated for the alloy 1430 (1.77 Li, 2.8 Mg, 1.7 Cu) sheets fabricated by hot and then cold rolling, quenching, and multistep aging showed a clear correlation between these factors and the real anisotropy of the strength properties of the sheets [42]. In this case, the main component of the matrix texture is $\{100\}\langle 011 \rangle$ with traces of $\{110\}\langle 112 \rangle$, and the surface layers also contain orientation $(111)[112]$.

The authors of [43] detected a significantly higher anisotropy of the strength properties of aluminum alloys with 25% lithium and 10% scandium as compared to other aluminum alloys in spite of the same textures of these alloys.

The formation texture in the δ' phase was systematically studied for the first time in [42, 44–47, 49], and the role of textured δ' in the anisotropy of the mechanical properties of Al–Li alloys was revealed. The textures of the solid solution and the δ' phase were analyzed in 7-mm-thick sheets and 25-mm-thick plates made of grade 1441 alloy in order to interpret the unusual shape of the fracture surfaces formed upon fatigue tests [45]. The following nonuniformity of the texture of the matrix phase over the cross sections of the sheets and plates was detected: the brass texture at the center and shear component $\{100\}\langle 011 \rangle$ in the surface layers. Recrystallization texture components $\{310\}\langle uvw \rangle$ were also detected. The character of texture of the δ' phase was different in the sheets and the plates: the texture of δ' in the sheets was analogous to the texture of the matrix phase, and that in the plates after heat treatment was mainly represented by the cube texture.

As was assumed in [45], the geometry of the fracture surfaces coincides with the arrangement of octahedral planes in the grains that belong to the texture component of brass. The difference between the textures of the matrix and the δ' phase is likely to be related to different intensities of the precipitation of this phase in matrix-phase grains of different orientations during cold deformation. The textures of the matrix and the δ' phase were studied in 2-mm-thick extruded section made of grade 1420 alloy subjected to quenching from 450°C followed by aging at 120°C for 1, 6, or 12 h [46]. The texture of the matrix phases was found to have the brass component and the cube component, which grew with aging time. The texture of the δ' phase only contained the cube component, which weakened with aging time because of the precipitation of a textureless phase.

The anisotropy of the yield strength in the Al–Li alloy sheet plane is associated with crystallographic texture, grain morphology, and oriented precipitates.

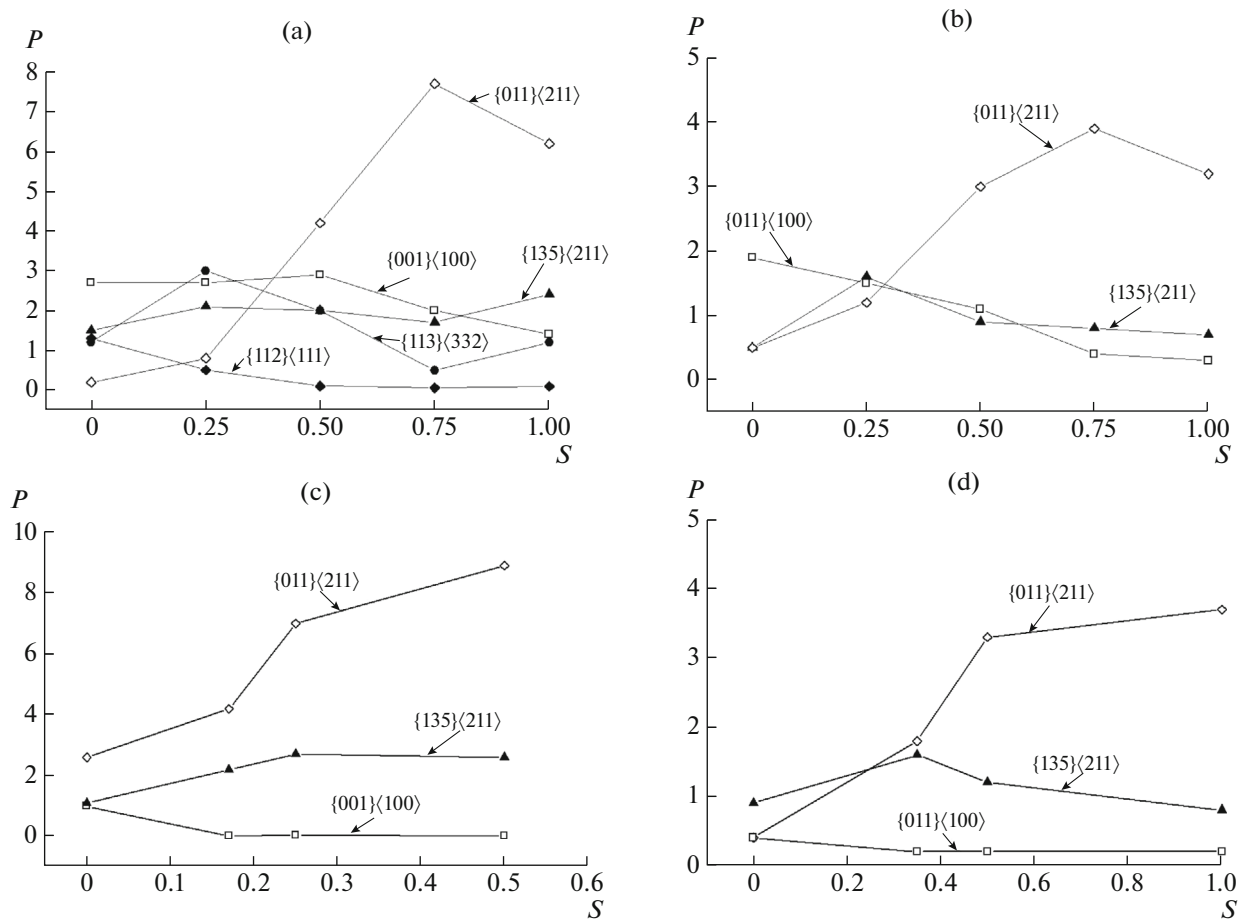


Fig. 10. Distributions of the texture components of phases (a, c) α and (b, d) δ' over the cross sections of sheets (a, b) 8 and (c, d) 1.6 mm thick. P is the pole density, $S = 2\Delta t/t$ is the relative distance from the sheet surface ($S = 1$ is the middle section) [48].

Hot-rolled Al–Li alloys have the minimum yield strength at an angle of 45° due to the formation of the brass and copper rolling textures and the presence of S component and elongated grains. Therefore, to decrease the anisotropy of the yield strength, it is necessary to weaken the rolling texture and to form equiaxed grains. This can be achieved by thermomechanical treatment, which includes overaging, warm rolling, and recrystallization treatment.

Fridlyander's works stimulated the development of quantitative approaches to explaining the anisotropy of the mechanical properties of Al–Li alloys, which is one of the challenging problems of these alloys. The texture and the anisotropy of the mechanical properties of alloy 1424 (5.6 Mg, 1.6 Li, 0.5 Zn, 0.1 Sc, 0.1 Zr) sheets produced by hot rolling (8-mm-thick sheets) and subsequent cold rolling to a thickness of 4 or 1.6 mm with intermediate annealing were studied in [48]. Texture was analyzed in various sections of the sheets, and their distance from the sheet surface was characterized by relative quantity $S = 2\Delta t/t$, where Δt

is the distance from the section to the surface and t is the sheet thickness. The investigations showed that the texture of the α solid solution is analogous to that of the δ' phase: they differ only in the intensity, which is usually higher in the solid solution. Brass component $\{011\}\langle 211 \rangle$ and weaker S component $\{135\}\langle 211 \rangle$ are predominant in the texture of the α solid solution in the central layer ($S = 0.5-1$) of the sheet (8-mm-thick) subjected to hot rolling, and component $\{113\}\langle 332 \rangle$ prevails in the subsurface ($S = 0.01-0.5$) layer (Fig. 10a). Cube texture component $\{001\}\langle 100 \rangle$ exists throughout the sheet thickness and its intensity decreases slightly toward the central layer. The change in the texture of the δ' phase over the cross section is similar to the change in the α solid solution: the central layer ($S = 0.01-0.5$) has two-component texture $\{011\}\langle 211 \rangle$ and $\{135\}\langle 211 \rangle$ with the predominant first component (Fig. 10b). Cold rolling slightly enhances brass texture $\{011\}\langle 211 \rangle$ for the α solid solution and extends it over the entire cross section of the sheet for both phases (Figs. 10c, 10d). The cube recrystallization texture component, which was detected over the

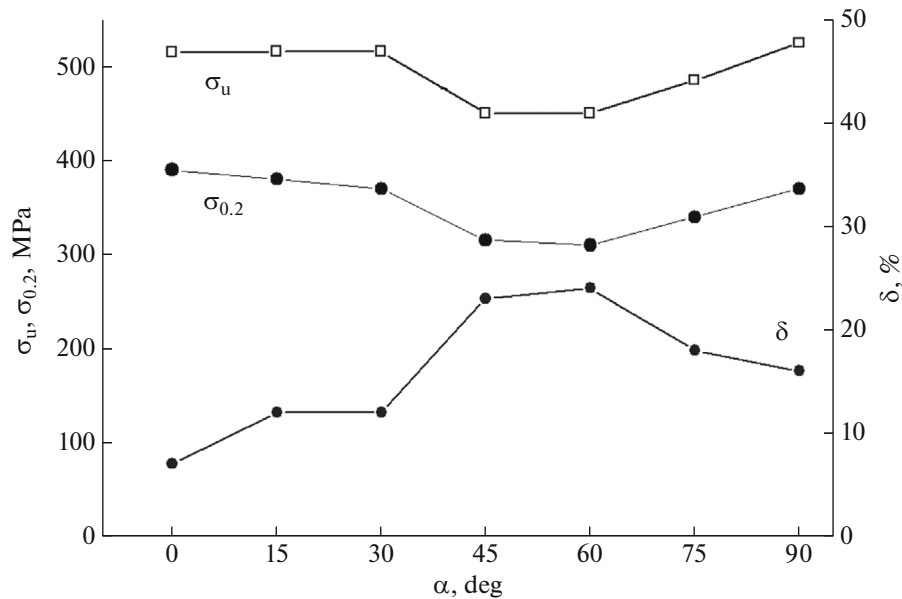


Fig. 11. Anisotropy of the mechanical properties of alloy 1424 sheets 8 mm thick. α is the angle to the rolling direction [48].

entire cross section of the hot-rolled sheet, is absent after cold rolling.

When generalizing these results, we would like to note the following two aspects. First, the intense texture of the δ' phase, which fully corresponds to the texture of the solid solution. (This fact was noted for the first time in [42, 49], where a relation between the textures of the δ' phase and the solid solution was experimentally detected.) The second important point is the absence of recrystallization in the sheets at all stages of rolling, including intermediate annealing and hardening heat treatment. This can be judged from the presence of intense rolling texture $\{011\}\langle 211 \rangle$, whereas the cube texture component, which points to recrystallization, is only present in the 8-mm-thick hot-rolled sheet. These findings indicate a strong joint effect of lithium and scandium on the deformation behavior of grade 1424 alloy, since the cube texture component in lithium-containing alloys without scandium is present in almost all sheets.

The alloys with lithium and scandium are characterized by a higher anisotropy of the mechanical properties than the alloys where these elements exist separately. We assume that the strong texture of the solid solution and δ' is caused by lithium and the fact that this texture does not disappear during recrystallization in the course of hot rolling and quenching is caused by scandium. Moreover, the effect of scandium is likely to manifest itself in the presence of the single-component brass texture, since copper texture component $\{112\}\langle 111 \rangle$ is usually important in the lithium-containing alloys.

The data shown in Fig. 11 demonstrate the anisotropy of the mechanical properties of hot-rolled 8-mm-thick sheets made of grade 1424 alloy. In particular, the yield strength of the sheets in RD and TD (transverse direction) is higher than that at an angle of 45° to the rolling direction (RD) by 21 and 18%, respectively. The strength properties of the sheets at an angle of 45° are low and the plasticity is high. Here, the anisotropy of the relative elongation exceeds 200%. Cold rolling weakly changes the anisotropy of the strength properties and significantly increases the anisotropy of the relative elongation, which is larger than 300 and 800% for 4- and 1.6-mm-thick sheets, respectively.

Inverse pole figures (IPFs) were obtained for three directions, namely, RD, 45° to RD, and TD, to interpret the anisotropy of the mechanical properties of the alloy 1424 sheets. As an example, Fig. 12 shows the IPFs of a 4-mm-thick sheet after cold rolling. It should be noted that the grain orientation in RD and TD (Figs. 12a, 12c) is more pronounced than that at an angle of 45° (Fig. 12b), which is caused by the symmetry of the texture components. This difference increases with the reduction during cold rolling. This finding should be taken into account to interpret the anisotropy of the properties, especially the anisotropy of plasticity.

The advantage of using IPF to calculate anisotropy in the case of a substantially nonuniform texture across the sheet thickness consists in the fact that the IPFs recorded from the cross section of the sheet give cross section—averaged information on texture. To

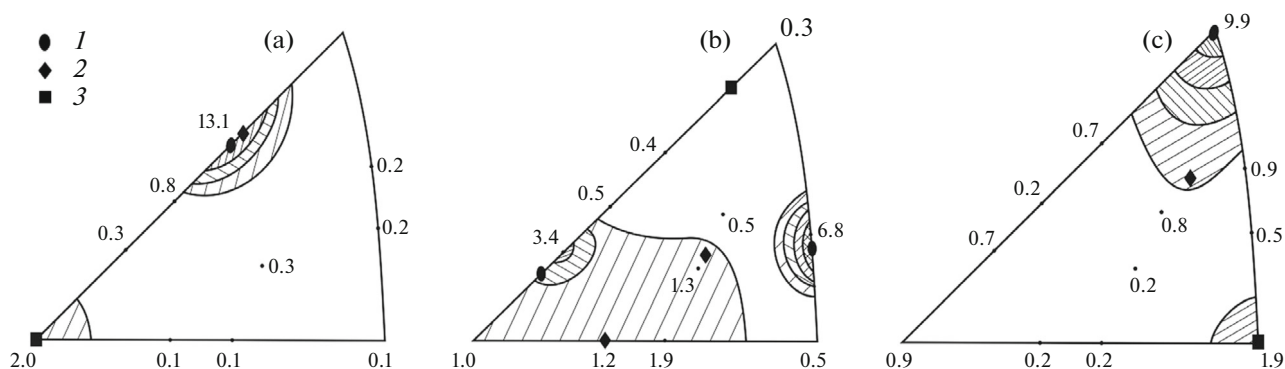


Fig. 12. IPF of a 4-mm-thick alloy 1424 sheet (MoK α radiation) [48]: (a) rolling direction (RD), (b) 45° to RD, and (c) transverse direction (TD). (1) {011}<211>, (2) {135}<211>, and (3) {011}<100>.

calculate the anisotropy of the strength properties, researchers use the following relationships:

$$\frac{\sigma_{0.2}^{\text{RD}}}{\sigma_{0.2}^{45}} = \frac{M^{\text{RD}}}{M^{45}} = \frac{\sum(\Phi_{hkl} P_{hkl})^{45}}{\sum(\Phi_{hkl} P_{hkl})^{\text{RD}}},$$

$$\frac{\sigma_{0.2}^{\text{TD}}}{\sigma_{0.2}^{45}} = \frac{M^{\text{TD}}}{M^{45}} = \frac{\sum(\Phi_{hkl} P_{hkl})^{45}}{\sum(\Phi_{hkl} P_{hkl})^{\text{TD}}},$$

where Φ_{hkl} is the Schmid factor for slip on the {111}<110> system for the loading axis that is normal to plane (hkl) and P_{hkl} is the pole density of reflection (hkl) on the IPF for the directions making an angle of 0°, 45°, or 90° with the rolling direction (RD, 45°, TD, respectively).

As the experimental values, the calculated anisotropies evidence low flow resistance in the 45° direction. However, in contrast to the experimental values, the flow resistance calculated in the 90° direction is higher than that in the rolling direction. This is natural, since the “strongest” direction in the fcc lattice (<111>) is located in the 90° direction (see Fig. 12c).

The difference between the experimental and calculated anisotropies was also noted in [43], where the effect of texture on the anisotropy of the mechanical properties of alloy 2195 (4.3 Cu, 1.4 Li, 0.4 Ag, 0.35 Mg, 0.13 Zr) sheets was studied. The anisotropy of the strength properties after quenching was shown to be 5–10%. Cold rolling by 24% in RD, TD, and 45° to the direction of preliminary rolling increased the anisotropy of the properties by 20–27%, with the anisotropy of the yield strength increasing after rolling in the 45° and 90° directions. It should be noted that the texture of the material changed insignificantly and its contribution to the anisotropy estimated from the Taylor factors was 8–12%. This value corresponds to the real anisotropy of the properties of the sheets after quenching and is significantly lower than the anisotropy of the cold-rolled sheets.

The anisotropy of the mechanical properties depends on the precipitation of the textured δ' phase during cold rolling. To explain this effect, it is neces-

sary to estimate the effect of the $L1_2$ -ordered δ' phase on the anisotropy of the alloy. This phase has a texture that is similar to that of the solid solution, and the phase deformation mechanism should have specific features related to a long-range order. For example, the maximum strength of the solid solution corresponds to orientation <111> and its minimum strength, to <100>. However, the maximum strength of the $L1_2$ structure corresponds to <100> (as in the case of nickel superalloy single crystals [50]), since the Schmid factor is zero for slip on cube planes {011}<110> for this orientation. For $L1_2$ -ordered structures, slip on the cube planes can be more preferential than slip on the close-packed planes at a high SFE, which is characteristic of aluminum alloys. The effect of additional silver and zinc alloying on the deformation characteristics of lithium-containing alloys is mainly related to the influence of these elements on SFE: it is likely to decrease, as in copper-based alloys.

An important specific feature of the alloys with lithium and scandium is the formation of shear bands during cold rolling, which is attributed to the existence of $L1_2$ -ordered δ' (Al_3Li) and Al_3Sc phases coherent with the aluminum matrix [51, 52]. The cutting of particles by paired dislocations causes deformation localization and, hence, the formation of shear bands. The factors affecting the formation of shear bands are as follows: the alloying element (lithium, scandium, magnesium) content; the state of alloy before cold rolling which determines the volume fraction of precipitates; and the cold rolling intensity.

The results obtained in [48] agree completely with the proposed mechanisms of shear band formation and can be used to refine them. In particular, the formation of coherent $L1_2$ -ordered Al_3Li - and Al_3Sc -phase precipitates ensures the maximum matching of the slip planes and directions in the {111}<110> system in the α and δ' phases only if slip proceeds in this system in the ordered δ' phase. (This orientation is most favorable for a dislocation to cross an interphase interface even provided that a perfect dislocation splits into

two partials with the formation of a stacking fault fringe.) In contrast, if slip in the δ' phase occurs in the $\{001\}\langle 110 \rangle$ system, the coherency between its particles and the matrix leads to the maximum discrepancy between slip planes in the α and δ' phases. In this case, deformation localization is inevitable and results in shear bands. Shear bands were found to affect the maximum strain per pass during cold rolling, the “reverse” anisotropy effect, and the thermal stability (DNN effect). The formation of shear bands is also associated with geometric softening $dM/de < 0$, where M is the Sachs factor [53].

6. CONCLUSIONS

(1) When analyzing the results of studying the formation of phase composition and texture in Al–Cu–Li and Al–Mg–Li alloys during deformation and heat treatment, we revealed quantitative relations to describe the effect of various intermetallic phases on the mechanical properties and their anisotropy of these alloys.

(2) Based on experimental measurements of the lattice parameter of the α solid solution, we developed a technique to calculate the amounts of the δ' , T_1 , and S_1 phases. The relation between the δ' phase and the ternary phases in Al–Cu–Li (Al–Mg–Li) alloys was shown to be determined by the atomic fractions of lithium and copper (magnesium).

(3) The calculations performed with the developed technique showed that the fraction of the δ' phase in Al–Cu–Li alloys is significantly higher than that of the ternary T_1 phase and that the fractions of δ' and the ternary S_1 phase in Al–Mg–Li alloys are approximately the same.

(4) The role of the T_1 phase as the main hardening component of Al–Cu–Li alloys is obviously overestimated. This is especially true of alloys with a lithium content more than 1.5%: the hardening effect due to the precipitation of the $L1_2$ -ordered δ' phase is predominant in these alloys. The hardening effect caused by the precipitation of ordered δ' -phase particles is proportional to the particle size, which ensures hardening upon aging even at a constant fraction of these particles.

(5) A multicomponent texture, which has predominant brass texture component $\{110\}\langle 112 \rangle$ and is non-uniform over the cross section of a sheet, forms in sheet semiproducts made of Al–Cu–Li and Al–Mg–Li alloys.

(6) The anisotropy of the mechanical properties of lithium-containing alloys is significantly higher than that of other aluminum alloys, which is explained by the presence of a significant content (up to 20%) of the coherent ordered δ' phase. The texture of this phase almost completely coincides with the texture of the solid solution, and the deformation mechanisms in this phase and the solid solution are radically different.

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