

Article

Comprehensive Analysis of Microstructure and Hot Deformation Behavior of Al-Cu-Y-Mg-Cr-Zr-Ti-Fe-Si Alloy

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Abstract: Low sensitivity to hot cracking is very important not only for casting but also for ingots of wrought alloys. Doping of Al-Cu-(Mg) alloys by eutectic forming elements provides an increasing resistance to hot cracking susceptibility, but it also leads to a decrease in plasticity. The quasi-binary alloys based on an Al-Cu-REM system with an atomic ratio of Cu/REM = 4 have a high solidus temperature, narrow solidification range and fine microstructure. The detailed investigation of microstructure, precipitation and hot deformation behavior, and mechanical properties of novel Al-Cu-Y-Mg-Cr-Zr-Ti-Fe-Si alloy was performed in this study. The fine Al₈Cu₄Y, needle-shaped Al₁₁Cu₂Y₂Si₂, compact primary (Al,Ti)₈₄Cu_{6.4}Y_{4.3}Cr_{5.3} and Q (Al₈Cu₂Mg₈Si₆) phases were identified in the as-cast microstructure. Near-spherical coarse Al₃(Zr,Y) and fine Al₄₅Cr₇ precipitates with a size of 60 nm and 10 nm were formed after 3 h of solution treatment at 580 °C. S'(Al₂CuMg) precipitates with an average diameter of 140 nm, thickness of 6 nm and calculated volume fraction of 0.033 strengthened 36 HV during aging at 210 °C for 3 h. Three-dimensional hot processing maps demonstrated an excellent and stable deformation behavior at 440–540 °C and strain rates of 0.01–10 s^{−1}. The rolled sheets had a good combination of yield strength (313 MPa) and plasticity (10.8%) in the recrystallized at 580 °C, with water quenched and aged at 210 °C for a 3 h state. The main calculated effect in the yield strength was contributed by Al₄₅Cr₇ precipitates.

Keywords: aluminum alloys; yttrium; phase composition; precipitates behavior; 3D processing maps; mechanical properties



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1. Introduction

Al-Cu-(Mg)-based alloys have a high strength and good heat resistance, but the worst casting properties in the Al alloys [1–5]. For example, commercial wrought 2024 (Al-Cu-Mg-Mn) alloy has a yield strength (YS) of 290–455 MPa after different aging treatments [2]. The YS decreases to about 130 MPa and 60 MPa after tensile tests at 205 °C and 260 °C [2]. The improved castability, for example, from low sensitivity to the hot cracking, is very important not only for casting but also for wrought alloys ingots [3–5]. Alloying of the Al-Cu-(Mg) alloys with Si, Ni, Fe, Mn, Ca leads to a decrease in hot cracking susceptibility, but an increased fraction of the brittle intermetallic compounds decreases the plasticity [4–9]. On the other hand, the search for novel alloying systems is the prospective aim. The alloys with atomic Cu/REM (rare earth metal) rotation of four in the Al-Cu-REM (REM are Ce,Y,Er) ternary system have a high solidus temperature, and as a result, a narrow solidification range [10–17]. Based on these qualities, the novel cast and wrought heat-resistant alloys were developed [18,19]. The two main rules were developed during alloy making: (1)—atomic Cu/REM ratio is equal to four; (2)—additional alloying by precipitates forms the elements Zr and Mn. Zirconium and manganese are well known as Al₃Zr [20–23]

and $\text{Al}_{20}\text{Cu}_2\text{Mn}_3$ [1,24,25] precipitates which form elements which are effective after high temperature annealing. Yttrium coupled with zirconium improves the strengthening since it increases the precipitates' density and the thermal stability of the $\text{L}_{12}\text{-Al}_3(\text{Zr,Y})$ [26–32]. Yttrium has two equivalent roles of eutectic phase and precipitate formation. The $\text{Al}_8\text{Cu}_4\text{Y}$ phase is defined by fine eutectic particles with high thermal stability which do not grow during solution treatment [12,13]. $\text{L}_{12}\text{-Al}_3(\text{Zr,Y})$ precipitates improve the strength and recrystallization behavior of the Al-Cu-Y-based alloys [28]. However, the addition of Mn to the Al-Cu-Y-based alloy provides a $\text{Al}_{25}\text{Cu}_4\text{Mn}_2\text{Y}$ phase solidification which does not dissolve during solution treatment [29]. As a result, the Mn leads to an increase of the fraction of intermetallic compounds in the solidification origin and copper content in the aluminum solid solution (Al). As a result, the Al-Cu-Y-based alloys with Mn have a good strength but low plasticity [18,19]. In this case, decreasing the Mn content or fully replacing it can help save the strength and improve the elongation. The Cr may also hold potential as an alloying element [1,33]. For example, the addition of Mn, Cr and Zr to the Al-Si-Mg-Cu alloy improves the creep behavior due to nano-sized $\alpha\text{-Al}(\text{Mn,Cr,Fe})\text{Si}$ and Al-Si-Zr dispersoid formation [33]. The Al-Cr-Si dispersoids' strengthening of an Al-Si-Mg-Cu casting alloy was studied in [34]. The chromium addition refines the grain structure, increases the yield strength and plasticity of the AA6061 alloy [35]. Scandium, chromium and zirconium increase the strength and ductility of the Al-Mg-Si-Mn alloy [36]. The $\text{Al}_{18}\text{Mg}_3\text{Cr}_2$ phase precipitates affect the heterogeneous nucleation sites for η phase during aging of the Al-Zn-Mg-Cu alloy [37]. The $(\text{Al,Cr})_3(\text{Zr,Yb})$ dispersoids were identified in the Al-Zr-Yb-Cr alloy after annealing at 500 °C [38]. The $(\text{Al,Cr})_3(\text{Zr,Yb})$ dispersoids are finer and more coarsening-resistant than the L_{12} -structured Al_3Zr and $\text{Al}_3(\text{Zr,Yb})$ dispersoids [38]. Based on the previous investigation the Al-Cu-Y-Mg-Mn-Zr-Ti-Fe-Si alloy, [18,19] was modified by replacing Mn with Cr.

The aim of the present study is to investigate the microstructure, phase composition, mechanical properties, hot deformation and recrystallization behavior of the new Al-4.8Cu-1.4Y-0.8Mg-0.2Cr-0.2Zr-0.15Ti-0.15Fe-0.15Si alloy. The composition of the novel alloy is close to the commercial 2024. In this case, the possible applications may be the same: aircraft structures, rivets, hardware, truck wheels, screw machine products and other miscellaneous structural applications [2].

2. Materials and Methods

2.1. Alloy Melting and Hot Cracking Index Determination

The content of alloying elements of the alloy is presented in Table 1. The alloy was melted in the resistance furnace Nabertherm at a temperature of 780 °C. Pure Al (99.7%), Cu (99.5%), Mg (99.5%) and Al-11Y, Al-10Cr, Al-5Ti-1B and Al-5Zr master alloys were used as raw materials. The melt was poured at a temperature of 780 °C into the water-cooled mold. The ingots' dimension was 120 × 40 × 20 mm. The cooling rate during solidification was about 15 °C/s. The chemical composition was measured using an electron diffraction X-ray (EDX) detector XMax-80 (Oxford Instruments Advanced AZtecEnergy, High Wycombe, UK) in a scanning electron microscope (SEM) Tescan Vega 3LMH (Tescan, Brno, Kohoutovice, Czech Republic). The hot cracking index (HCI) was measured using the pencil probe technique [1,3–5]. The pencil probe is a rapid and simple method for HCI determination. Four pourings were performed to determine the average HCI value.

Table 1. Chemical composition of the investigated alloy, wt.%.

Al	Cu	Y	Mg	Zr	Cr	Fe, Si, Ti
bal.	4.8	1.4	0.8	0.2	0.2	0.15 of each

2.2. Processing Maps Construction

Processing maps were created based on compression tests at elevated temperatures after solution treatment using a Gleeble 3800 (Dynamic Systems Inc., Poestenkill, NY, USA).

Specimens with a diameter of 10 mm and a height of 15 mm were compressed. Compression tests were processed at temperatures of 400–540 °C and at strain rates of 0.01–10 s^{−1}. Friction and adiabatic heating during deformation were considered to recalculate the compression stress–strain curves [39,40]. The OriginLab Software (Version 9.1, Northampton, MA, USA) was used to process maps' data with B-spline interpolation.

2.3. Materials Processing and Mechanical Properties Measurements

An ingot treated with solution treated at 580 °C for 3 h, with a thickness of 20 mm, was water quenched and hot rolled at 540 °C to a thickness of 8 mm (1 mm reduction per pass) and at 25 °C to a 1 mm (0.5 mm reduction per pass) thickness sheet. The scheme for the production and heat treatment of the sheets is presented in Figure 1. The structure and mechanical properties of rolled and annealed specimens were investigated. The hardness was measured with the standard Vickers method under a 5 kg load. The tensile specimens were tested on a Zwick/Roell Z250 Allround (Zwick/Roell, Kennesaw, GA, USA) test machine. The tensile specimens with a gauge length of 22 mm and gauge width of 6 mm were cut out from 1 mm thickness sheets in the rolling direction. Three specimens were tested in one state.

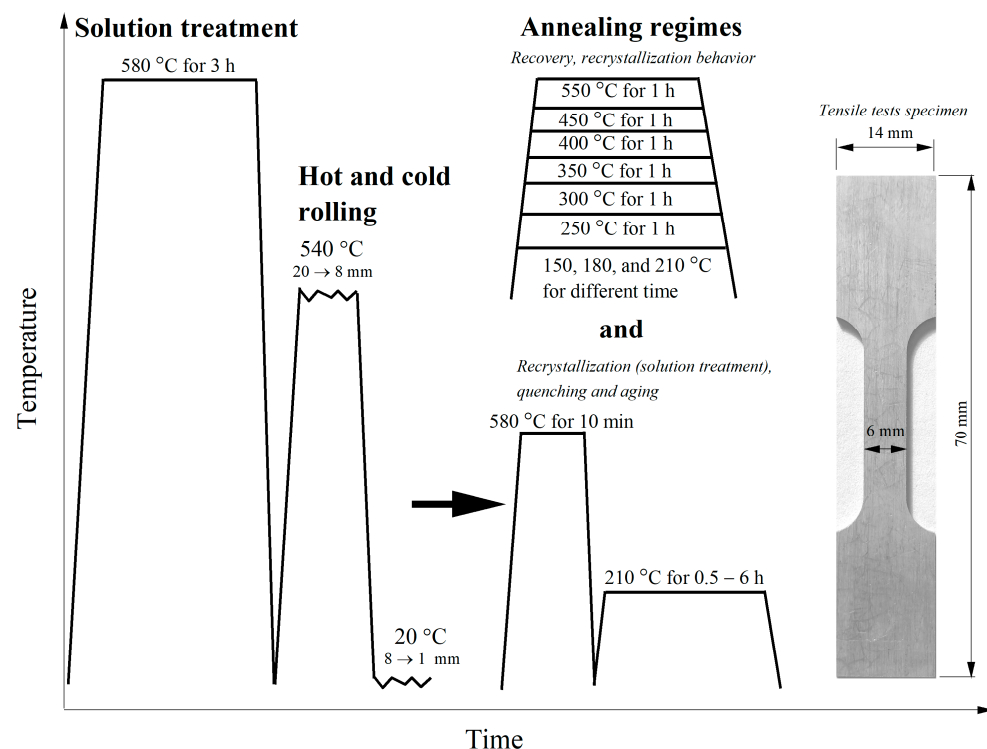


Figure 1. Regimes of the thermo-mechanical treatment.

2.4. Microstructure Characterization

The microstructure and phase composition were investigated in detailed by SEM. The precipitates were studied after solution and aging treatment in a transmission electron microscope (TEM) JEOL-2100 (Jeol Ltd., Tokyo, Japan). An optical microscope (OM) Carl Zeiss Axiovert 200 (Carl Zeiss AG, Oberkochen, Germany) was used to investigate grain microstructure after annealing the rolled sheets.

The solidus temperature was determined in a differential scanning calorimeter (DSC) Labsys Setaram (SETARAM Instrumentation, Caluire, France). The solution treatment, annealing of the rolled sheets and aging treatment were carried out in resistance furnaces with fans. The specimens for SEM and OM investigations were mechanically grinded and polished with Struers Labobol equipment (Struers APS, Ballerup, Denmark). The specimens for grain microstructure investigation were electrochemically etched in a Barker's

solution (46 mL of HBF_4 , 7 g of HBO_3 , and 970 mL of H_2O) at a voltage of 15–25 V and temperature of 0–5 °C. The TEM specimens were electrochemically polished in A2 electrolyte on Struers Tenupol-5 equipment (Struers APS, Ballerup, Denmark). Thermodynamic calculations of the multicomponent phase diagram and phase equilibria were carried out in the Thermo-Calc software (TCW5, Thermo-Calc Software AB, Stockholm, Sweden) in the TTAL5 database.

3. Results and Discussion

3.1. As-Cast Microstructure

The aluminum solid solution (Al) and four different intermetallic phases are clearly divided in the as-cast microstructure (Figure 2). The main structure component is ((Al)+ $\text{Al}_8\text{Cu}_4\text{Y}$) eutectic (spectrum 1 in Figure 2). The Fe impurity was completely dissolved in the $\text{Al}_8\text{Cu}_4\text{Y}$ phase (spectrum 1 in Figure 2). The same results were obtained in our previous investigations [18,19,41]. The fine particles of the $\text{Al}_8\text{Cu}_4\text{Y}$ phase effectively increases the yield strength of the Al-Cu-Mg alloy, as was demonstrated by the authors of [42]. Brighter needle-shaped $\text{Al}_{11}\text{Cu}_2\text{Y}_2\text{Si}_2$ phase particles with lengths of 5–8 μm and a low volume fraction were also identified (spectrum 2 in Figure 2). The same phase particles were found in the Al-Cu-Y alloy with Fe and Si impurities [41]. Chromium in combination with Y, Cu, and Ti leads to the primary coarse $(\text{Al,Ti})_{84}\text{Cu}_{6.4}\text{Y}_{4.3}\text{Cr}_{5.3}$ phase formation with a size of 10 μm (spectrum 3 in Figure 2). The similar $\text{Al}_{81-85}\text{Cu}_{7-10}\text{Y}_{3-4}\text{Cr}_5$ phase was firstly identified in the AlCuYZrCr alloy without Ti [43]. The Ti may substitute Al or Y atoms in the lattice of this phase. Copper and magnesium provide the solidification of the Q ($\text{Al}_8\text{Cu}_2\text{Mg}_8\text{Si}_6$) phase (spectrum 4 in Figure 2). Copper, magnesium, zirconium, yttrium and chromium content in the (Al) in the as-cast state is 1.2, 0.5, 0.3, 0.2 and 0.2, respectively (Table 2). The solidus temperature of 591 °C was determined by DSC. The solution treatment temperature of 580 °C was chosen in accordance with the solidus temperature.

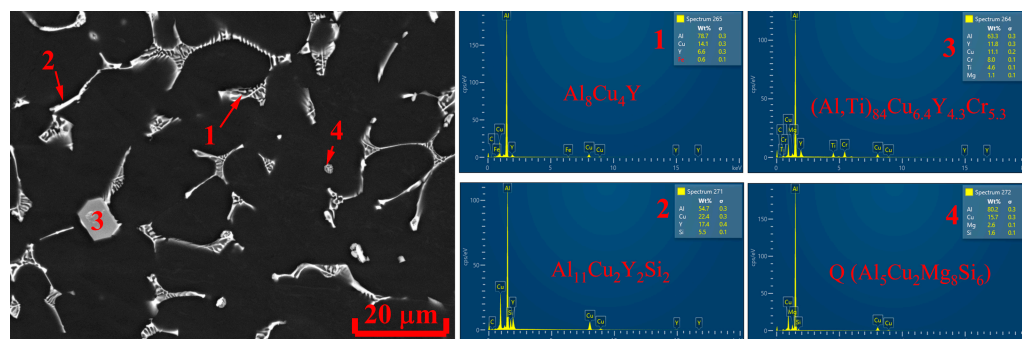


Figure 2. As-cast microstructure and EDS spectra from intermetallic particles (SEM): point 1— $\text{Al}_8\text{Cu}_4\text{Y}$ phase, point 2— $\text{Al}_{11}\text{Cu}_2\text{Y}_2\text{Si}_2$ phase, point 3— $(\text{Al,Ti})_{84}\text{Cu}_{6.4}\text{Y}_{4.3}\text{Cr}_{5.3}$ phase, point 4—Q ($\text{Al}_8\text{Cu}_2\text{Mg}_8\text{Si}_6$) phase.

Table 2. Composition of the (Al) in wt.% (EDX SEM).

State	Al	Cu	Mg	Y	Zr	Cr
as-cast	bal.	1.2	0.5	0.2	0.3	0.2–0.3
580 °C for 3 h	bal.	1.9	1.0	0.2	0.3	0.2–0.3
580 °C for 6 h	bal.	2.1	1.0	0.2	0.3	0.2–0.3

3.2. Solution-Treated, Quenched and Aged Microstructure

Two diffusion-controlled processes occur during solution treatment: homogenization and heterogenization. Dissolution of the non-equilibrium phases and fragmentation, spheroidization and growth of the equilibrium phases are the main processes during homogenization. The zinc, magnesium and copper content in the (Al) increased due to the dissolvment of the non-equilibrium phases, such as the Q-phase (Table 2). The microstruc-

tures of the solution-treated alloy at 580 °C for 3 h and 6 h are presented in Figure 3. The main eutectic $\text{Al}_3\text{Cu}_4\text{Y}$ phase fragmented, spheroidized and grew from a thickness of less than 200 nm (Figure 2) to a diameter of 1000 nm (Figure 3a) after 3 h of solution treatment. The most equilibrium form with the lowest energy is spherical. The thin and elongated particles fragmented to take on a nearly spherical shape. The fragmentation process is controlled by the diffusion of copper and an yttrium atom through the aluminum matrix. Increasing the solution treatment time to 6 h did not affect the size of the intermetallic phases of the solidifications' origin. The needle-shaped $\text{Al}_{11}\text{Cu}_2\text{Y}_2\text{Si}_2$ phase particles are clearly seen after solution treatment and did not change the morphology (Figure 3).

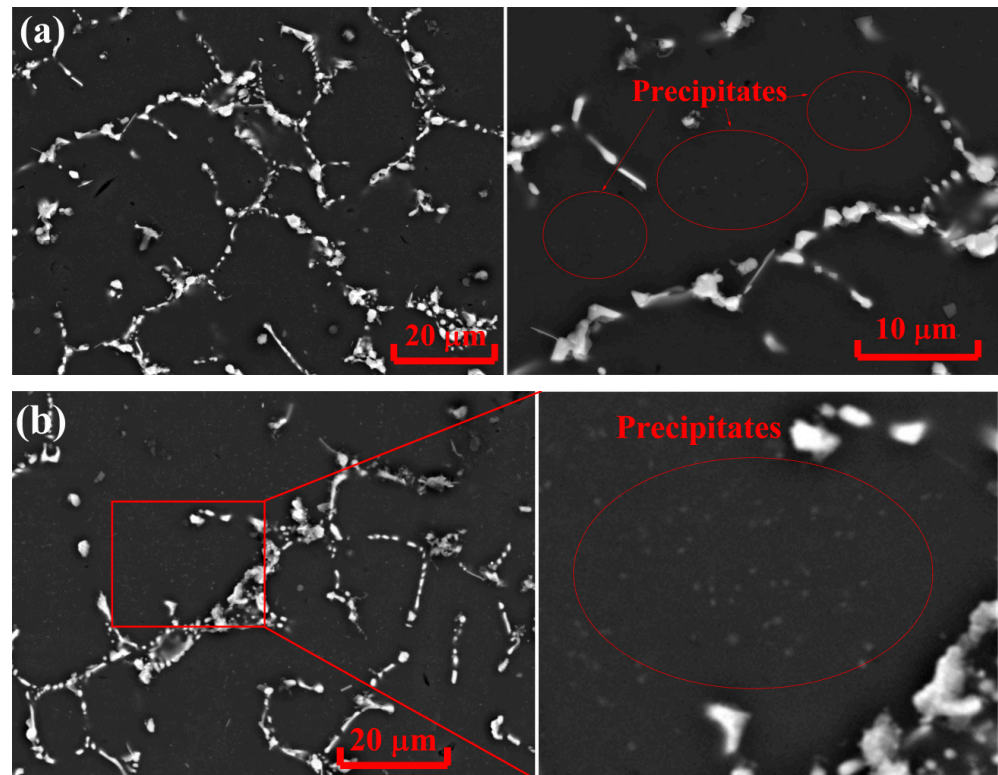


Figure 3. The microstructure of the solution-treated alloy at 580 °C for (a) 3 h and (b) 6 h (SEM).

The parallel process with homogenization during solution treatment is the heterogenization with nucleation of the precipitates from supersaturated (Al) by Zr, Y and Cr after solidification. The calculated polythermal section $\text{Al-2Cu-1Mg-0.3Zr-(0-0.4)Cr}$ was constructed to analyze the phase composition of the (Al) after decomposition (Figure 4). The $\text{Al}_{45}\text{Cr}_7$ and $\text{D}_{023}\text{-Al}_3\text{Zr}$ phases must be in equilibrium with (Al) during solution treatment at a temperature of 580 °C. The fine particles are clearly seen in the (Al) after 3 h of solution treatment (Figure 3a). The size of these particles is significantly increased after 6 h of solution treatment (Figure 3b). Based on the stabilization of the Cu and Mg content (Table 2) and finer precipitates, 3 h was chosen as an optimal solution treatment time.

The investigated alloy was water quenched and aged at 210 °C for 3 h, after 3 h of solution treatment at 580 °C. The detailed TEM results of the precipitates are presented in Figure 5. The alloy was aged at 150–210 °C to determine the regime of maximum strengthening (Figure 5). Three types of precipitates were identified in the microstructure (Figure 5). Near-spherical coarse $\text{Al}_3(\text{Zr,Y})$ and fine $\text{Al}_{45}\text{Cr}_7$ (6 nm) precipitates were nucleated during the solution treatment, and disk-shaped $\text{S}'(\text{Al}_2\text{CuMg})$ precipitates were nucleated during the aging treatment. The EDX-TEM point analysis demonstrates the presence of Zr and Y in the near-spherical coarse particles. The $\text{Al}_3(\text{Zr,Y})$ particle structure was determined as D_{023} according to the fast Fourier transformation (FFT) in Figure 5. The calculated FFT of $\text{D}_{023}\text{-Al}_3(\text{Zr,Y})$ is the same as the electron diffraction of $\text{D}_{023}\text{-Al}_3\text{Zr}$ [44].

However, the Moiré effect in the left side of the particle in Figure 5c indicates the presence of a partially coherent boundary. It means that the $L1_2 \rightarrow D0_{23}$ transformation after 3 h of solution treatment was completed in most parts. The $L1_2 \rightarrow D0_{23}$ transformation of the Al_3Zr -based precipitates were analyzed in detail [23,45,46]. The anti-phase boundary has an important role in this transformation [23,45,46]. At the same time, the $Al_3(Zr,Y)$ precipitates in the Al-Cu-Y-Zr-Cr alloy without Mg after 3 h of annealing at 600 °C have an $L1_2$ structure with anti-phase boundaries and a smaller diameter of 50 nm [43]. The average size of the $D0_{23}$ - Al_3Zr precipitates in the investigated alloy is 60 nm. The Mg content of 1% in the (Al) accelerates the $Al_3(Zr,Y)$ precipitates' growth and the $L1_2 \rightarrow D0_{23}$ transformation at 580 °C. For example, Mg solute atoms in a concentration of 1–5% has a negligible influence on the precipitation of $Al_3(Er,Zr)$ after 3 h of annealing at 150–550 °C [47].

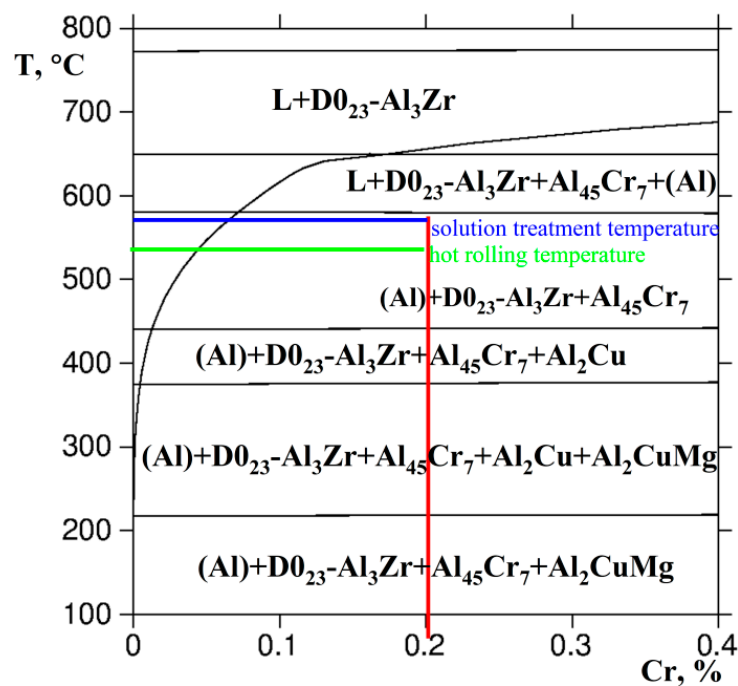


Figure 4. The calculated polythermal section of the multicomponent Al-2Cu-1Mg-0.3Zr-(0–0.4)Cr system.

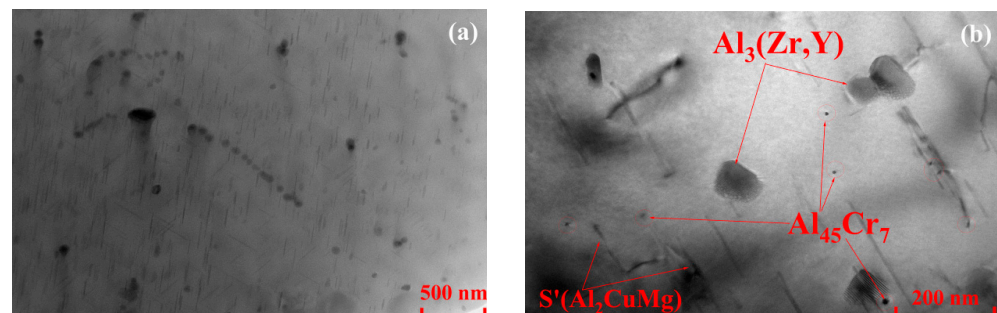


Figure 5. Cont.

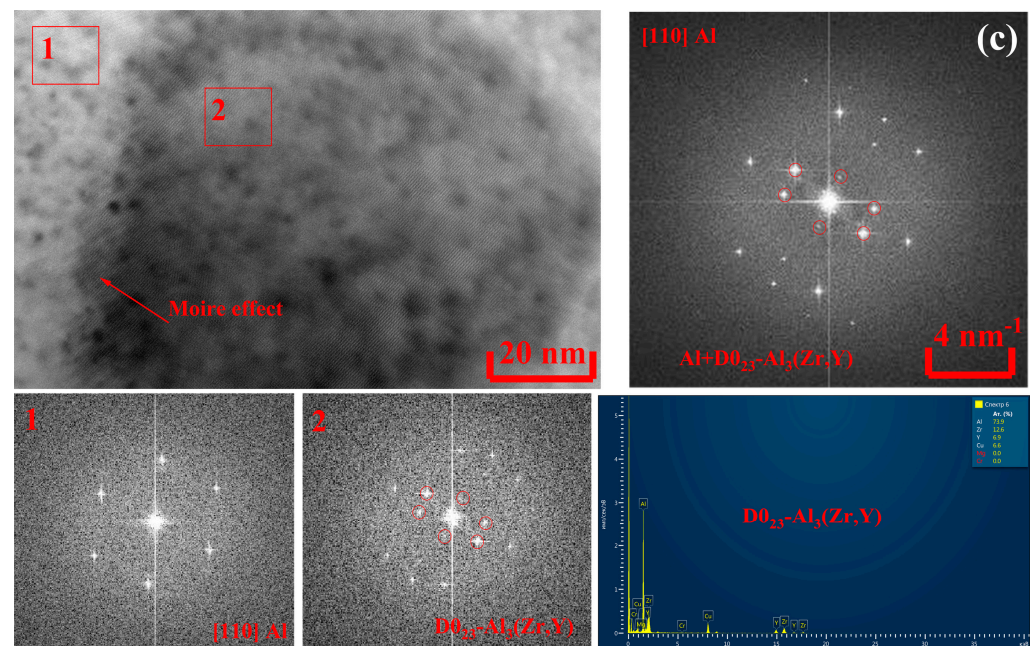


Figure 5. The microstructure of the water-quenched and aged alloy at 210 °C for 3 h after 3 h of solution treatment at 580 °C at (a–c) different magnification (inserts in (c)—FFT from TEM image, (1) Al matrix, and (2) precipitate and EDS spectrum from precipitates).

3.3. Hot Deformation Behavior

The stress–strain curves are shown in Figure 6. The temperature increases enhanced dislocation mobility, which accelerated the annihilation and rearrangement of dislocations. The strain rate increases should offer a stronger interaction between dislocation and a decreased time for dynamic softening. As a result, the stress increased as the compression temperature decreased and the strain rate increased.

The construction of hot processing maps (HPM) is an extremely useful approach for determining the optimal regime for hot deformation. These maps are created based on the dynamic material Prasad and Gegel model [48]. Calculating the efficiency of power dissipation during deformation is a crucial step in forming the processing map. By employing this method, the optimal deformation regimes have been successfully determined for numerous aluminum alloys [49–59]. Detailed methodologies for creating HPMs have been described in previous investigations [49–64]. The power dissipation efficiency η was calculated using the equation [50]:

$$\eta = \frac{J}{J_{\max}} = \frac{2m}{m+1}, \quad (1)$$

where J is the power consumed by the microstructure evolution, such as dynamic recrystallization, recovery and phase transformations; $J_{\max} = (\sigma\dot{\epsilon})/2 = P/2$; and m is the strain rate sensitivity coefficient.

The 3D HPMs for the investigated alloy are presented in Figure 7. The highest η value was achieved by increasing the temperature and decreasing the strain rate. However, over the entire strain range, the alloy demonstrates the highest η at temperatures of 440–540 °C and strain rates of 0.01–1 s^{−1}.

In addition, the flow stability of the material during hot deformation (the absence of the potential places for fracture) was calculated as the dimensionless parameter of ξ [49]:

$$\xi(\dot{\epsilon}) = \left[\frac{\partial \ln \left(\frac{m}{m+1} \right)}{\partial \ln \dot{\epsilon}} \right] + m, \quad (2)$$

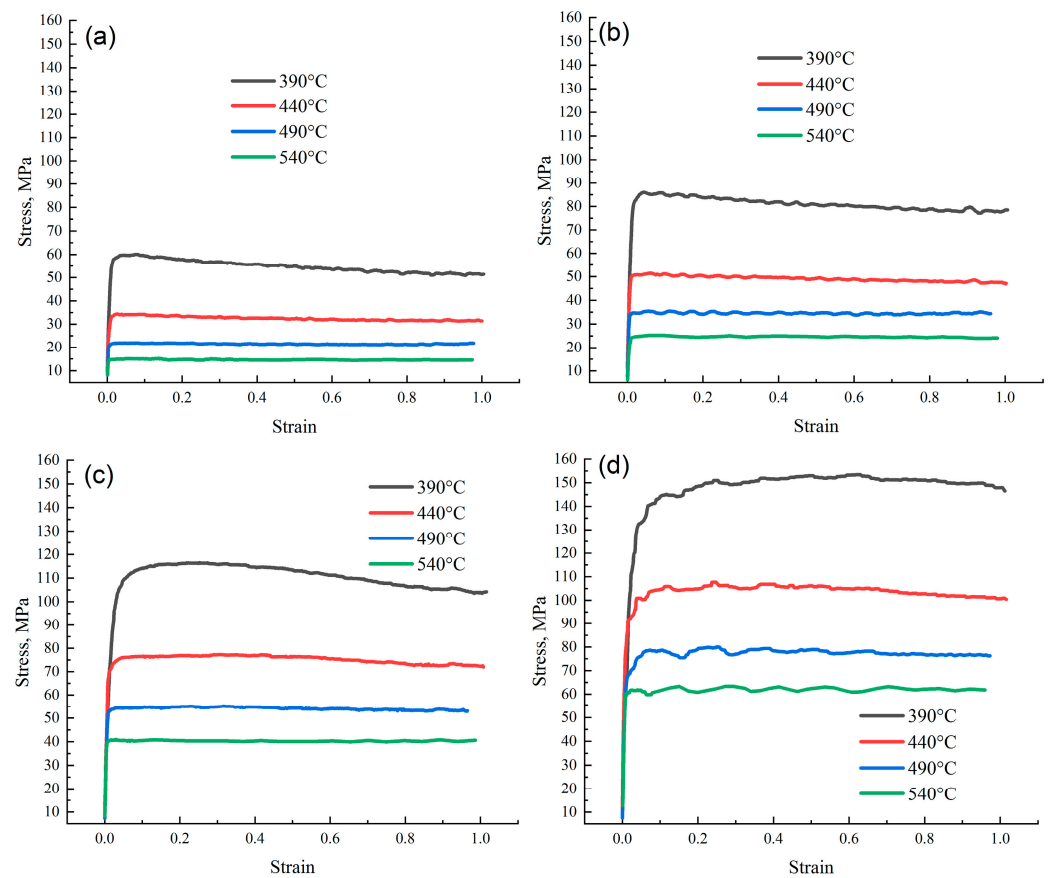


Figure 6. True stress–strain curves for the investigated alloy deformed at strain rates of (a) 0.01 s^{-1} , (b) 0.1 s^{-1} , (c) 1 s^{-1} and (d) 10 s^{-1} at temperatures of 390–540 °C.

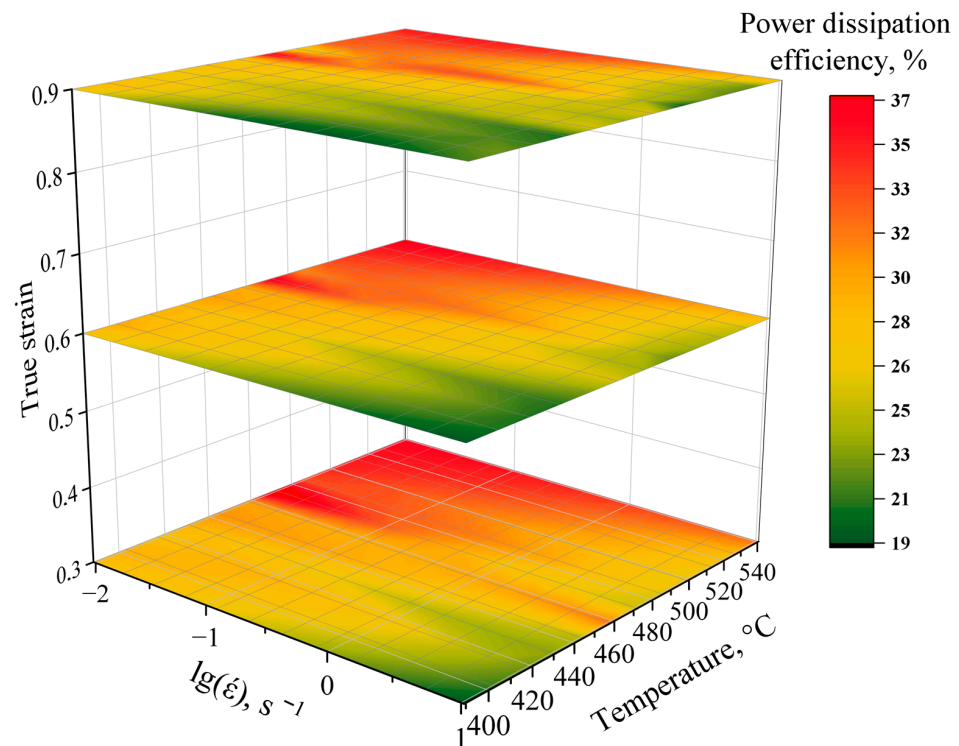


Figure 7. 3D HPM of the investigated alloy—the power dissipation efficiency η is dependent on the temperature, strain and strain rate.

The values of the instability criteria are positive in all ranges of the deformation temperatures, strains and strain rates; this means that there is a good hot deformation flow of the investigated alloys in the chosen range of hot deformation parameters.

3.4. Recovery, Recrystallization Behavior and Mechanical Properties

3.4.1. Recovery and Recrystallization

The specimens of heavy cold-rolled sheets were annealed at 100–550 °C for 1 h, and the HV and grain structure evaluation was investigated (Figure 8). The hardness in the as-rolled state is 127 HV. Softening and strengthening are the two processes which occur during annealing of rolled sheets at temperatures of 100–150 °C. Softening occurs due to recovery processes or decreasing the dislocation density and substructure with low angular grain boundaries' formation. The $\text{Al}_3(\text{ZrY})$ and $\text{Al}_{45}\text{Cr}_7$ precipitates are effectively suppressing the recovery. The second parallel process is strengthening due to aging. The hot rolling at high temperature and cooling on air provide the quenched structure. In this case, the strengthening from aging prevails over the recovery softening. The same results were obtained in the other Al-Cu-Mg-based alloys [15]. The aging effect disappears with the increase of the annealing temperature to 300 °C. The softening gradually accelerates, but the microstructure is still non-recrystallized. A completely recrystallized structure with a grain size of 6 μm was achieved after 1 h of annealing at 400 °C (Figure 8). Increasing the annealing temperature to 450 °C did not significantly affect the grain size but the HV increased from 65 to 75. The Al-Cu-Mg-based alloys are very sensitive to nature aging. The hardness measurements after annealing were performed after a long time. The increasing annealing temperature from 400 to 450 °C promotes the formation of a more saturated solid solution and higher hardening due to nature aging. The same results were indicated earlier [19].

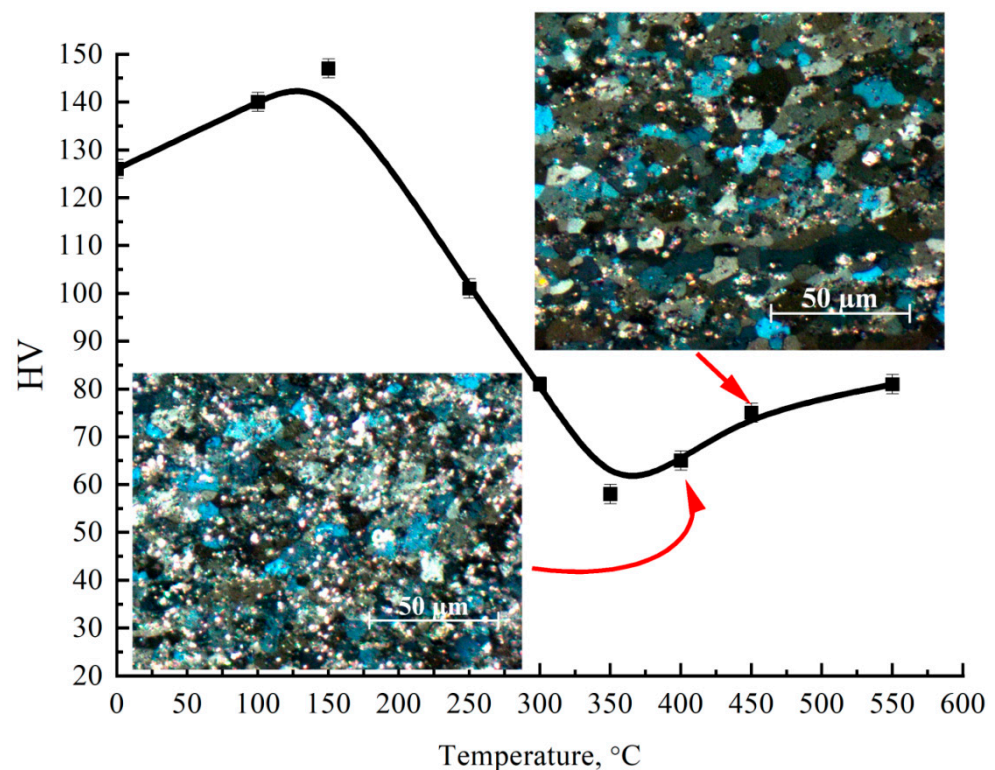


Figure 8. HV temperature dependencies for 1 h of annealing after rolling (inserts—the grain structure after 1 h of annealing at 400 and 450 °C).

3.4.2. Mechanical Properties after Low Temperature Annealing

HV time dependencies demonstrate in detail the competition between recovery and aging. The aging completely prevails over the recovery softening during annealing at 150 °C, and the alloy has highest yield strength (YS) of 404 MPa at a good elongation of 3%. The aging dominates over softening up 1 h of annealing at 180 °C (Figure 9). The YS = 354 MPa was obtained after 3 h of annealing at 180 °C. The recovery softening during annealing at 210 °C completely covers the aging hardening, and the alloy has the lowest yield strength (YS) of 317 MPa at a good elongation of 3.2%.

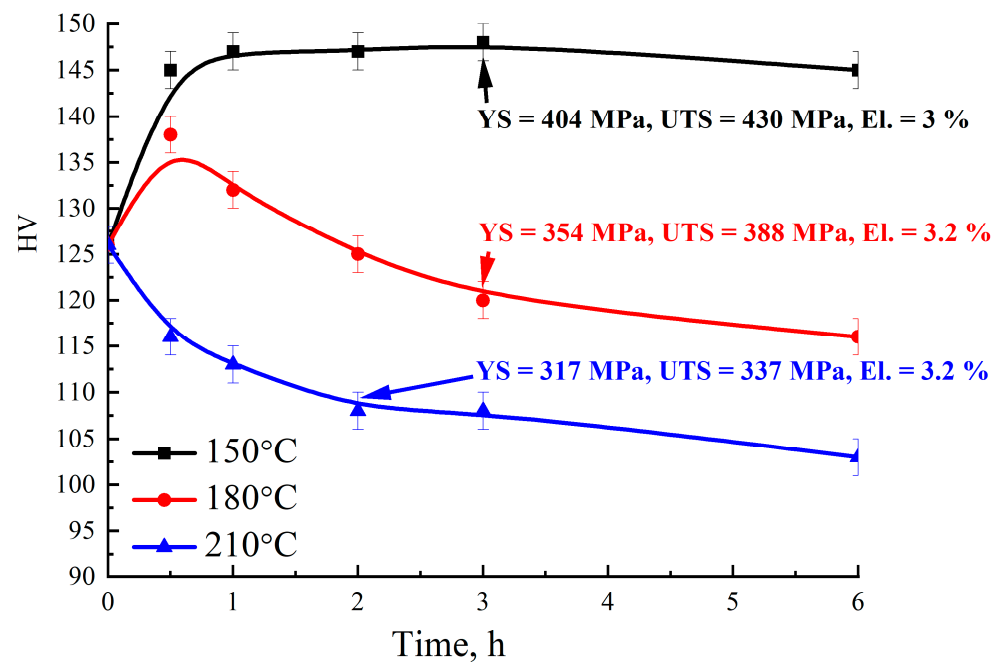


Figure 9. HV time dependencies of the sheets annealed at 150–210 °C.

3.4.3. Hardness Variation with Aging

The low temperature annealing cannot provide a sufficiently high plasticity in most cases. This is why the heat-treatable alloys are subjected to recrystallization annealing at a solution treatment temperature, quenching and aging. The alloy sheet was recrystallized at 580 °C for 10 min, water quenched and aged at 210 °C (Figure 10). The hardness of the quenched sample was measured immediately. As-quenched hardness is 69 HV. The average grain size is 10 µm. The maximum hardness of 119 HV during aging was achieved after 3 h (Figure 10). The investigated alloy demonstrates an excellent combination of strength and plasticity: YS = 313 MPa, UTS = 390 MPa and El. = 10.8% in this state. The 2024 alloy is a commercial alloy closest in composition to the Al-Cu-Mg system [2]. 2024 alloy sheets under T6 have a lower YS at the same UTS: YS = 287 MPa, UTS = 395 MPa and El. = 18.9% [65]. The absence of solidification origin coarse particles in the 2024 alloy sheets provides a higher elongation than in the investigated alloy. The addition of the precipitates forming Sc and Zr elements in the 2024 alloy increased the YS to 313 MPa, but it decreased the elongation to 16.7% [65]. However, Sc is the most expensive alloying element in the Al alloys. In addition, the Al-4Cu-1.5 Mg alloy (2024 alloy base elements [2]) has the highest sensitivity to the hot cracking formation due to a wide solidification range [3,4]. The investigated alloy has a significantly lower hot cracking susceptibility. The HCI = 12–14 mm of the investigated alloy is similar to the cast Al-Si-Cu-Mg alloys of the 3 series [4,5].

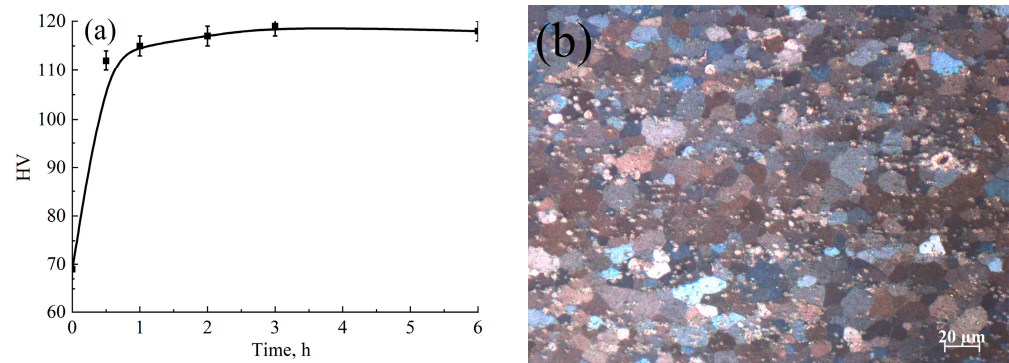


Figure 10. (a) changes in hardness with aging at 210 °C for various durations after recrystallization at 580 °C for 10 min and quenching, and (b) the grain structure of the alloy sheet annealed at 580 °C for 10 min.

3.5. Strengthening Mechanisms Contribution

Five strengthening mechanisms were calculated based on structure parameters and calculated volume fraction of the precipitates. The contributions from grain boundaries ($\Delta\sigma_{gb}$), dislocations ($\Delta\sigma_d$), solid solution ($\Delta\sigma_{ss}$), precipitates ($\Delta\sigma_{ppt}$) and solidification origin particles $\Delta\sigma_p$ were calculated with the following equations [66–73]:

$$\Delta\sigma_{gb} = kd^{-0.5}, \quad (3)$$

$$\Delta\sigma_d = M\alpha_1 Gb\sqrt{\rho_{dis}}, \quad (4)$$

$$\Delta\sigma_{ss} = 57.5C_{Cu} + 13.6C_{Mg}, \quad (5)$$

where $k = 0.065 \text{ MPa}/\text{m}^{-2}$ is the Hall–Petch slope, d is the average grain diameter; $M = 3$ is the mean orientation factor or Taylor factor, $\alpha_1 = 0.3$ is the frequency-centered cubic lattice constant of Al, $G = 26 \text{ GPa}$ is the Shear modulus of Al, $b = 0.286 \text{ nm}$ is Burger’s vector, $\rho_{dis} = 10^9 \text{ sm}^{-2}$ is the dislocation density, where C_{Cu} and C_{Mg} are the atomic concentration of Cu and Mg in the (Al);

$$\Delta\sigma_{ppt}^{Or.sphere} = \frac{M \cdot 0.4 \cdot Gb}{\pi \sqrt{(1-v)}} \cdot \frac{\ln\left(\frac{2\bar{R}}{r_0}\right)}{\lambda}, \quad (6)$$

$$\Delta\sigma_{ppt}^{Or.disk} = 0.13 \left(\frac{M G b}{\sqrt{d \cdot w}} \right) \left(\sqrt{f} + 0.75 \cdot \sqrt{\frac{d}{w}} \cdot f + 0.14 \cdot \frac{d}{w} \cdot f^{3/2} \right) \cdot \left(\ln \left(\frac{0.87 \cdot \sqrt{d \cdot w}}{r_0} \right) \right) \quad (7)$$

where $r_0 = 1.5 b$; $v = 0.345$ is the Poisson’s ratio; $\bar{R} = \frac{\pi R_s}{4}$ is the mean planar radius (where R_s is precipitate radius); $\lambda = R_s \cdot \left(\sqrt{\frac{2\pi}{3f}} - \frac{\pi}{4} \right)$ is the edge-to-edge interprecipitate spacing; d —precipitate diameter; w —precipitate thickness; f —volume fraction of the precipitates or particles.

Equation (6) was used to calculate the contribution from spherical (or near-spherical) solidification origin particles and precipitates. Equation (7) was used to calculate the contribution from disk-shaped precipitates of the aging origin. Following superposition ($\sigma_0 = 10 \text{ MPa}$ is the friction stress) was taken in the calculation model of the yield strength of the alloy:

$$\sigma_y = \sigma_0 + \Delta\sigma_{gb} + \Delta\sigma_{ss} + \Delta\sigma_p + \sqrt{\Delta\sigma_d^2 + \Delta\sigma_{ppt}^2} \quad (8)$$

The results of the calculated contribution from structure components are summarized in Table 3. The volume fractions of the $\text{Al}_3(\text{Zr,Y})$, $\text{Al}_{45}\text{Cr}_7$ and S' precipitates were calculated from binary Al-Zr and Al-Y and quaternary Al-Cu-Mg-Zr-Cr phase diagrams. Zirconium and yttrium content in the (Al) are about 0.3 and 0.2 wt%, respectively (Table 2). The radii of grains, solidification origin particles and precipitates were measured from OM, SEM and TEM images. The calculated yield strength (σ_y) of 307.8 MPa is very close to the experimental YS = 313 MPa (Figure 9). The main calculated effect in the yield strength was contributed from $\text{Al}_{45}\text{Cr}_7$ precipitates.

Table 3. Calculated contribution from different structure components.

Strengthening Mechanism	Structure Parameters	Contribution, MPa
$\Delta\sigma_{gb}$	$k = 0.065 \text{ MPa/m}^{-2}$, $d = 10 \text{ }\mu\text{m}$ (Figure 9)	20.5
$\Delta\sigma_d$	$\rho_{dis} = 10^9 \text{ sm}^{-2}$ [5]	21.2
$\Delta\sigma_{ss}$	$C_{Cu} = 0.1\%$, $C_{Mg} = 0.28\%$	6.7
$\Delta\sigma_{ppt}^{Or.sphere}$	eutectic particles ($r = 375 \text{ nm}$, $f = 0.1$ (Figure 2))	18.0
	$\text{Al}_3(\text{Zr,Y})$ ($r = 30 \text{ nm}$ (Figure 4), $f = 0.008$)	35.9
	$\text{Al}_{45}\text{Cr}_7$ ($r = 3 \text{ nm}$ (Figure 4), $f = 0.0084$)	188
$\Delta\sigma_{ppt}^{Or.disk}$	S' ($d = 140 \text{ nm}$, $h = 6 \text{ nm}$ (Figure 5), $f = 0.033$)	147
σ_y		307.8

4. Conclusions

The detailed investigation of the microstructure, phase composition, precipitation and hot deformation behavior and mechanical properties of the novel Al-Cu-Y-Mg-Cr-Zr-Ti-Fe-Si alloy was carried out by using OM, SEM, TEM, thermodynamic and 3D processing map calculation, and mechanical properties testing.

1. The fine $\text{Al}_8\text{Cu}_4\text{Y}$ phase with a thickness of less than 200 nm, the needle-shaped $\text{Al}_{11}\text{Cu}_2\text{Y}_2\text{Si}_2$ phase with a length of 5–8 μm and compact primary $(\text{Al,Ti})_{84}\text{Cu}_{6.4}\text{Y}_{4.3}\text{Cr}_{5.3}$ phase with a size of less than 10 μm and a very low volume fraction and the Q ($\text{Al}_8\text{Cu}_2\text{Mg}_8\text{Si}_6$) phase were identified in the as-cast microstructure.
2. Near-spherical coarse $\text{Al}_3(\text{Zr,Y})$ (60 nm) and fine $\text{Al}_{45}\text{Cr}_7$ (6 nm) precipitates were formed after 3 h of solution treatment at 580 °C. The Moire effect is present in part of the $\text{Al}_3(\text{Zr,Y})$ precipitate and indicates a partially coherent boundary. This indicates an almost complete $\text{L}_{12} \rightarrow \text{D}_{023}$ transformation after 3 h of solution treatment.
3. $S'(\text{Al}_2\text{CuMg})$ precipitates with average diameter of 140 nm, thickness of 6 nm and calculated volume fraction of 0.033 strengthen 36HV during aging at 210 °C for 3 h.
4. 3D hot processing maps demonstrate an excellent and stable deformation behavior at 440–540 °C and strain rates of 0.01–10 s^{-1} due to the uniform structure with coarse particles of the solidification origin and fine precipitates.
5. The rolled sheets have a good combination of yield strength (313 MPa) and plasticity (10.8%) in the recrystallized alloy at 580 °C, which had been water quenched and aged at 210 °C for 3 h state. The calculated yield strength (σ_y) of 307.8 MPa is very close to the experimental YS = 313 MPa. The main calculated effect in the yield strength was contributed from $\text{Al}_{45}\text{Cr}_7$ precipitates. 2024 alloy sheets under T6 have a lower YS at the same UTS: YS = 287 MPa, UTS = 395 MPa and El. = 18.9%. In this case, the possible applications may be the same: aircraft structures, rivets, hardware, truck wheels, screw machine products and other miscellaneous structural applications.

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