Effect of Zr on Microstructure and Mechanical Properties of the Al–Cu–Yb and Al–Cu–Gd Alloys

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Abstract: The effect of zirconium on the microstructure, phase composition, and mechanical properties of AlCuYb and AlCuGd alloys was studied. The microstructure of the as-cast alloys did not consist of new intermetallic phases of zirconium with other elements, so the zirconium was fully dissolved in the aluminum matrix. The AlCuYbZr/AlCuGdZr alloys demonstrated higher hardness values compared to the AlCuYb/AlCuGd alloys due to the precipitation of the Al_{3}(Zr,Yb) and Al_{3}(Zr,Gd) phases, which were formed during the homogenization treatment. The AlCuYbZr alloy had a 10–20 MPa higher yield and tensile strength than the AlCuGdZr alloy at the same annealing temperature and time. The AlCuYbZr alloy exhibited good mechanical tensile properties at an annealing temperature of 100 °C for 1 h, with a yield strength of 276 MPa, ultimate tensile strength of 312 MPa, and elongation of 3.1%, while the as-rolled AlCuGdZr alloy had similar mechanical tensile properties, with a yield strength of 279 MPa, ultimate tensile strength of 307 MPa, and elongation of 4.8%. At an annealing temperature of 300 °C for 10 min, the AlCuYbZr and AlCuGdZr alloys showed a good ductility of 10.5% and 8%, respectively, with 207 MPa yield strength for both alloys. AlCuYbZr and AlCuGdZr alloys are a prospective base composition for developing novel high technology heat resistant aluminum alloys.

Keywords: aluminum alloys; ytterbium; gadolinium; zirconium; mechanical properties

1. Introduction

One of the major aluminum alloy systems with precipitation hardening is the aluminum-copper systems with strengthening from CuAl2 [1]. The mechanical properties of Al–Cu alloys are highly dependent on the parameters of the precipitates, such as the lattice structure, number density, shape, and size and distribution of the particles [2]. Al–Cu alloys have a high strength at room and elevated temperatures [1,3]. The high heat resistance of the Al–Cu alloys provides thermal stability to the θ’ (Al2Cu) precipitates [2,3]. Cast Al–Cu alloys generally have a coarse grain structure, inhomogeneous intermetallic phases, and low castability. A coarse structure causes stress concentration and acts as premature crack initiation in tension. Improvement of the mechanical and casting properties can be achieved, for example, by heat and ultrasonic treatment, alloying, and grain refinement [4].

Alloying by rare earth metals (REM) is one way to improve the mechanical properties of aluminum alloys [4–32]. REMs are known to be favorable for alloying purification and degassing. Adding REM to the Al alloys can reduce the secondary dendrite arms spacing, refine grains, and affect the precipitation response [6]. The quasi-binary alloys of the Al–Cu–REM system have been investigated by several researchers, for example Al–Cu–Ce [7,8], Al–Cu–Y [9–11], and Al–Cu–Er [9,12,13], with an atomic ratio of 4/1 (Cu/REM) due to the formation of an Al6Cu4REM phase of a crystallization origin, which increases
the mechanical properties of the Al–Cu alloys. This alloying principle was used in the investigation of the ternary quasi-binary Al–4.4Cu–2.5Yb and Al–4.5Cu–2.5Gd alloys [14]. Recently, some studies have investigated the effect of zirconium on the microstructure and precipitation strengthening of Al–Cu–RE alloys. The addition of a zirconium in Al–Cu–Er alloys increases the recrystallization temperature and tensile strength due to the precipitation of Al$_3$(Er,Zr) [15–17]. The precipitation strengthening effect increases in the Al–Er–Zr alloys due to L1$_2$ precipitates of the Al$_3$(Er,Zr) phase [18–21]. The Al$_3$(Er,Zr) precipitates are similar to the Al$_3$(Y,Zr) that is formed in Al–Cu–Y–Zr and Al–Cu–Y–Mn–Zr alloys [22,23]. The novel cast and wrought alloys with improved castability, room temperature, and elevated temperature strength were developed based on the quasi-binary Al–Cu–Y and Al–Cu–Er alloys [24,25]. Gadolinium or ytterbium should be a promising couple for scandium [26–31] or zirconium [29–35] in order to achieve precipitate strengthening in aluminum and its alloys.

This work aims to present an analysis and evaluation of the microstructure and mechanical properties of new Al–Cu–Yb and Al–Cu–Gd alloys by adding Zr.

2. Materials and Methods

AlCuYbZr and AlCuGdZr alloys were prepared by casting metallurgy in an electric resistance furnace. Solidification was achieved in a copper water-cooled mold with a cavity size of $20 \times 40 \times 120$ mm$^3$, with a cooling rate of about 15 Ks$^{-1}$. The raw materials were pure Al (99.99% Al and 0.01% of Fe and Si) and Al–52Cu, Al–10Yb, Al–10Gd, and Al–5Zr master alloys. The content of the main elements in the alloys is presented in Table 1. The melt temperature before casting was 800 °C. The nominal composition of the alloys is shown in Table 1. The Labsys Setaram Differential Scanning Calorimeter (SETARAM Instrumentation, Caluire, France) (DSC) was used to determine the liquidus and solidus temperatures. The tests were carried out in an air atmosphere. In the investigated temperature range, the heating and cooling rate was 5 °C/min. X-ray diffraction (XRD) data were collected on a Bruker D8 Advance diffractometer with Cu–Kα radiation.

| Alloy   | Cu | Yb | Gd | Zr | Al   |
|---------|----|----|----|----|------|
| AlCuYbZr | 4.1| 2.2| -  | 0.4| bal. |
| AlCuGdZr | 4.2| -  | 2.2| 0.4| bal. |

Heat treatment was carried out in Nabertherm and SNOL furnaces with a fan and a temperature control accuracy of 1 °C. Homogenization of AlCuYbZr and AlCuGdZr alloys was carried out at 590 and 605 °C, respectively, for 1, 3, and 6 h. After the homogenization of AlCuYbZr and AlCuGdZr alloys at 590 and 605 °C, respectively, for 6 h, the samples were exposed to quenching, followed by aging at 150, 180, and 210 °C for 0.5–6 h. The other steps after homogenization were hot rolling to 10 mm at 440 °C, then cold rolling to 1 mm. The temperature of hot rolling of the ingots did not exceed the homogenization temperature. As-rolled alloys were annealed at 150, 180, and 210 °C for 0.5–6 h and at 100–550 °C for 1 h.

The samples for the microstructural analysis before and after deformation were prepared using a Struers LaboPol-5 polishing machine (Struers APS, Ballerup, Denmark), including the stages of mechanical grinding and polishing. The surface was ground on sandpaper of different grain sizes (220, 320, 800, 1200, 2400, and 4000). Then, microsections were subjected to mechanical polishing using an OP-S suspension. To reveal the grain structure after electropolishing, we used anodic oxidation in a 10% aqueous solution of hydrofluoric acid at a voltage of 16–20 V. Oxidation was carried out at a temperature of 10 °C. The oxidation time was 30–50 s. The grain sizes were measured from photomicrographs obtained in the polarized light of a light microscope (LM) using the random secant method. Structure analysis and phase identification were carried out on a Neophot
30 light microscope (LM) and on a TESCAN VEGA 3LMH scanning electron microscope (SEM) using an energy-dispersive X-ray (EDX) detector X-Max 80 and a Bruker D8 Advance X-ray diffractometer. A transmission electron microscope (TEM) JEOL JEM-2100 EX (Jeol Ltd., Tokyo, Japan) was used for the microstructure investigation on disc-shaped electrochemically-thinned samples with a diameter of 3 mm.

Tensile tests were carried out on a Zwick/Roell Z250 Allround testing machine (Zwick/Roell, Kennesaw, GA, USA) in combination with an automatic longitudinal strain sensor. The tensile rate during testing was 4 mm/min. The tested samples had a gage length of 20 mm, a width of 6 mm, and a thickness of 1 mm. The hardness was measured using the standard Vickers method. The hardness number (HV) was determined as the arithmetic mean of five measurements for annealed specimens and of ten measurements for the cast. The error in the determination did not exceed 3 HV.

3. Results and Discussion

The grain structure of the as-cast AlCuYbZr and AlCuGdZr alloys is presented in Figure 1. The average grain size of the AlCuYbZr and AlCuGdZr alloys was 230 and 240 ± 20 μm, respectively (Figure 1). The same grain structure was found in the ternary AlCuYb and AlCuGd alloys. The addition of Zr did not have an effect on the grain structure of the Al–Cu–Y alloys [10,22]. An average grain size of 190 ± 20 μm was found in the Al–Cu–Y–Zr alloy in comparison with 180 ± 30 μm in the Al–Cu–Y alloy. The microstructure of the as-cast AlCuYbZr and AlCuGdZr alloys is shown in Figure 2. Aluminum solid-solution, eutectic structure, and bright inclusions of Al₃Yb or (Al,Cu)₁₇Yb₂ phases are identified in the AlCuYbZr as-cast alloy (Figure 2a), while the as-cast microstructure of AlCuGdZr (Figure 2b) is represented by Al dendrites and a eutectic structure. The eutectic structure of AlCuYbZr/AlCuGdZr alloys consists of an aluminum solid-solution and Al₅Cu₄Yb/Al₅Cu₄Gd (τ₁) phase. Zirconium was fully dissolved in the aluminum solid solution (Al) as in the AlCuEr [15,16] and AlCuY [20] alloys.

The XRD patterns of the as-cast AlCuYbZr and AlCuGdZr alloys in comparison with the Zr-free alloys are shown in Figure 3 [14]. The X-ray diffraction analysis of AlCuYbZr alloy in Figure 3a reveals peaks for aluminum, Al₅Cu₄Yb, and small peaks of the Al₃Yb or (Al,Cu)₁₇Yb₂ phases. The XRD of the as-cast AlCuGdZr alloy in Figure 3b shows peaks for the aluminum and Al₅Cu₄Gd phase.

![Figure 1](image1.png)

Figure 1. As-cast grain structure of the AlCuYbZr (a) and AlCuGdZr (b) alloys (LM, polarized light).
The microstructure after homogenization at different times before quenching at 590 and 600 °C, respectively. The microstructure after homogenization was examined by DSC. Based on the DSC results, the homogenization temperatures for AlCuYbZr and AlCuGdZr alloys were 590 and 600 °C, respectively.

The solidus temperatures of the AlCuYbZr (Figure 4a) and AlCuGdZr (Figure 4b) alloys were 603 and 613 °C, respectively, as examined by DSC. Based on the DSC results, the homogenization temperatures for AlCuYbZr and AlCuGdZr alloys were 590 and 600 °C, respectively.
The AlCuYbZr and AlCuGdZr alloys were homogenized at different times before quenching at 590 and 600 °C, respectively. The microstructure after homogenization treatment is shown in Figure 5. The size of the intermetallic particles and copper concentration in the (Al) as a function of homogenization time at 590 and 600 °C are presented in Figure 6. The homogenization annealing for 1 h led to a fragmentation and spheroidization and growth of the Al$_8$Cu$_4$Yb and Al$_8$Cu$_4$Gd eutectic phases; the mean size increased from 0.25 µm (Figures 5a,b and 6) to 1.9 and 2 µm in the AlCuYbZr and AlCuGdZr alloys, respectively. In this case, the Cu concentration in the (Al) increased from 1.2–1.3% in the as-cast state to 2.1 and 1.6% after 6 h of annealing in the AlCuYbZr and AlCuGdZr alloys, respectively (Figure 6). By increasing the homogenization time, the second phases grew and the concentration of copper in (Al) increased. The increased copper concentration was as a result of dissolving the non-equilibrium phases of a solidification origin. However, it was observed that the copper concentration of 2.1% in (Al) in the AlCuYbZr alloy (Figure 6) slightly exceeded the corresponding value for the AlCuGdZr alloy (1.6%) after 6h of homogenization. Aging after quenching at 150–210 °C did not provide a hardening effect due to the very low copper content in the (Al).

|  | 1h | 3h | 6h |
|---|---|---|---|
| AlCuYbZr alloy | ![Image](a) | ![Image](b) | ![Image](c) |
| | ![Image](20µm) | ![Image](20µm) | ![Image](20µm) |
| AlCuGdZr alloy | ![Image](d) | ![Image](e) | ![Image](f) |
| | ![Image](20µm) | ![Image](20µm) | ![Image](20µm) |

**Figure 5.** Microstructure evaluation of the AlCuYbZr (a–c) and AlCuGdZr (d–f) alloys after homogenization treatment for 1 (a,d), 3 (b,e), and 6 (c,f) h.
The TEM microstructure investigation after 6 h of homogenization, quenching, and aging at 210 °C for 3 h was applied (Figure 7). The spherical and disk-shaped particles were present in the investigated alloys. The point chemical analysis demonstrated that the spherical particles were enriched in Zr and Yb in the AlCuYbZr alloy, and Zr and Gd in the AlCuYbGd alloy. The disk-shaped particles were enriched in Cu. Spherical particles typical for L1\textsubscript{2} structured dispersoids with a composition of Al\textsubscript{3}(Zr,Yb) and Al\textsubscript{3}(Zr,Gd) were nucleated during homogenization treatment. Disk-shaped particles are typical for the θ'(Al\textsubscript{2}Cu) phase, which forms during aging treatment. The size of the Al\textsubscript{3}(Zr,Yb) and Al\textsubscript{3}(Zr,Gd) phase dispersoids was in the range of 42–50 nm in diameter. The discs of the θ'(Al\textsubscript{2}Cu) phase were 80–180 nm in diameter and 3–5 nm in thickness. The disc of the θ'(Al\textsubscript{2}Cu) phase was nucleated at the boundary of the Al\textsubscript{3}(Zr,Yb) and Al\textsubscript{3}(Zr,Gd) phase dispersoids and the aluminium solid solution. The same structure was formed in the Al–Cu–Y and Al–Cu–Er alloys with Zr and Mn additions [17,23].

The ingots were rolled into 1 mm thick sheets after homogenization for 6 h. The as-rolled specimens were annealed at 150, 180, and 210 °C for 0.5–6 h. The hardness value decreased with increasing the annealing time and temperature for both alloys (Figure 8a,b). After annealing the AlCuYbZr/AlCuYb alloys at 180 and 210 °C for 6 h, a slight decrease in hardness of 14–22 HV was observed in the Zr-bearing alloy, while the decrease in the Zr-free alloy was 22–32 HV (Figure 8a,b). On the other hand, during annealing at 150, 180, and 210 °C for 6 h for the AlCuGdZr/AlCuGd alloys, a slight decrease in hardness of 11–20 HV was observed in the Zr-bearing alloys, while the decrease in the Zr-free alloys was 23–43 HV (Figure 8b). Thus, the softening was significantly lower in the studied alloys compared to the similar Zr-free alloys, which could be attributed to the formation of Al\textsubscript{3}(Zr,Yb) and Al\textsubscript{3}(Zr,Gd) precipitates, which provided Zener pinning and Orovan strengthening mechanisms.
Figure 7. Microstructure (TEM) of the AlCuYbZr (a) and AlCuGdZr (b) alloys after 6 h of homogenization, quenching, and aging at 210 °C for 3 h.

Figure 8b,c demonstrates the hardness HV vs. temperature dependences of AlCuYbZr/AlCuYb and AlCuGdZr/AlCuGd alloys. The AlCuYbZr/AlCuGdZr alloys showed the same hardness vs. temperature, but a higher hardness value compared to the Zr-free alloys due to the formation of Al3(Zr,Yb) and Al3(Zr,Gd) phase precipitates. The hardness of the AlCuYbZr deformed alloys decreased with increasing the temperature up to 350 °C to 54 HV due to the static recovery and recrystallization processes. However, when temperature exceeded 350 °C, the hardness slightly increased to 57 and 60 HV at 450 and 550 °C, respectively. This could be attributed to supersaturated (Al) formation, accompanied by repeated hardening of the alloying solution [36]. The authors prove that there may be another reason for the increase in hardness above 350 °C for the AlCuYbZr alloy, because the hardness of AlCuYb (Zr-free) increased above 450 °C. The presence of the Al3(Zr,Yb) phase could have a role in increasing the hardness over 350 °C on the other hand. The hardness of the AlCuGdZr deformed alloys decreased with increasing the temperature up to 450 °C to 39 HV and slightly increased to 45 HV at 550 °C.

Figure 9 shows the grain structure of the AlCuYbZr and AlCuGdZr alloys annealed at 350–550 °C after rolling. The average grain size of the AlCuGdZr alloy was 8 μm after annealing at 450 °C and slightly increased to 8.5 μm after annealing at 550 °C, while the grain size in the AlCuGdZr alloy increased from 10 μm to 11 μm in the same condition.

Table 2 summarized the YS, UTS, and elongation of the studied alloys after rolling and annealing. The AlCuYbZr alloy had a higher yield (YS) and ultimate tensile strength (UTS) than the AlCuGdZr alloy at the same annealing temperature and time due to a higher copper content in (Al) after homogenization treatment. The AlCuYbZr alloy exhibited a good YS of 276 MPa, UTS of 312 MPa, and elongation to failure of 3.1% at an annealing temperature of 100 °C for 1 h, while the as-rolled AlCuGdZr alloy had a good YS of 279 MPa, UTS of 307 MPa, and elongation to failure of 4.8%. The AlCuYbZr and AlCuGdZr alloys possessed a good ductility of 10.5% and 8%, respectively, at an annealing temperature of 300 °C for 10 min with YS of 207 MPa for both alloys. The AlCuYbZr and AlCuGdZr alloys
had higher mechanical tensile properties than AlCuYb and AlCuGd [14] after annealing at temperatures above 150 °C due to the thermal stability of these alloys, which were obtained by adding Zr.

Figure 8. HV vs. time (a,b) and annealing temperature (1 h) (c,d) dependencies of the rolled AlCuYb/AlCuYbZr (a,c) and AlCuGd/AlCuGdZr (b,d).

Table 2. Tensile properties of the AlCuYbZr and AlCuGdZr alloys in the as rolled and annealed states in comparison with Zr-free alloys.

| State            | YS, MPa | UTS, MPa | El., % | YS, MPa | UTS, MPa | El., % |
|------------------|---------|----------|--------|---------|----------|--------|
|                  | AlCuYbZr |          |        | AlCuGdZr |          |        |
| As-rolled *      | 288 ± 2  | 325 ± 2  | 2.5 ± 0.5 | 279 ± 1  | 309 ± 3  | 4.8 ± 0.4 |
| 100 °C for 1 h   | 276 ± 3  | 312 ± 3  | 3.1 ± 0.4 | 268 ± 2  | 302 ± 2  | 4.1 ± 1.0 |
| 100 °C for 6 h   | 272 ± 4  | 307 ± 4  | 2.7 ± 0.3 | 258 ± 3  | 286 ± 4  | 4.6 ± 0.6 |
| 150 °C for 1 h   | 265 ± 1  | 295 ± 1  | 4.1 ± 1.3 | 255 ± 2  | 284 ± 3  | 3.8 ± 0.5 |
| 150 °C for 6 h   | 262 ± 2  | 287 ± 3  | 2.6 ± 1.0 | 251 ± 2  | 277 ± 2  | 1.8 ± 1.0 |
| 180 °C for 1 h   | 254 ± 4  | 281 ± 2  | 5.2 ± 0.3 | 248 ± 2  | 272 ± 3  | 2.5 ± 0.8 |
| 180 °C for 6 h   | 252 ± 1  | 274 ± 2  | 3.5 ± 1.0 | 237 ± 2  | 261 ± 2  | 2.5 ± 1.1 |
| 210 °C for 1 h   | 252 ± 2  | 272 ± 2  | 1.9 ± 0.4 | 237 ± 1  | 260 ± 2  | 3.7 ± 1.2 |
Table 2. Cont.

| State                  | YS, MPa | UTS, MPa | EL, % | YS, MPa | UTS, MPa | EL, % |
|------------------------|---------|----------|-------|---------|----------|-------|
| 250 °C for 0.5 h       | 236 ± 2 | 252 ± 1  | 4.5 ± 0.5 | 225 ± 5 | 235 ± 4  | 5.2 ± 0.4 |
| 300 °C for 10 min      | 207 ± 2 | 226 ± 2  | 10.5 ± 1 | 207 ± 3 | 212 ± 2  | 8 ± 1 |

| AlCuYbZr [14]          |         |          |       |         |          |       |
| 210 °C for 1 h         | 238 ± 2 | 253 ± 1  | 5.9 ± 0.1 | 202 ± 1 | 210 ± 1  | 6.7 ± 0.7 |
| 250 °C for 0.5 h       | 216 ± 2 | 229 ± 1  | 6 ± 2  | 175 ± 4 | 182 ± 3  | 16.0 ± 1.2 |

* period of the storage of the samples prior to testing was more than 1 week.

Figure 9. Grain structure of the AlCuYbZr (a–c) and AlCuGdZr (d–f) alloys after rolling and subsequent annealing at 350 (a,d), 450 (b,e), and 550 (c,f) °C for 1 h.

The strengthening effect of the L12 precipitates were calculated. The precipitates’ contribution to the mechanical properties was provided by precipitate shearing or precipitate bypass looping mechanisms [37]. L12 precipitates with a radius greater than the threshold of 2.4 nm to 3.1 nm were non-shearable [37,38] and contributed to YS via the bypass looping mechanism, which can be estimated using the Orowan equation [37,39–41]

\[
\Delta \sigma_{or} = \frac{M \cdot 0.4 \cdot Gb}{\pi \sqrt{(1 - v)}} \frac{\ln \left( \frac{2 \pi R_s}{r_0} \right)}{\lambda},
\]

where \( M \) is approximately 3.0 and is the Taylor factor, \( G = 26 \) GPa is the shear modulus for Al, \( b = 0.286 \) nm is the Burgers vector for Al, \( \bar{R} = \frac{n R_s}{4} \) is the mean planar radius of the precipitates, \( R_s \) is the precipitate radius determined from the TEM images of \( r_0 = 1.5b \), and
\( v = 0.345 \) is the Al Poisson’s ratio \([37,39]\). The edge-to-edge interprecipitate spacing \( \lambda \) is calculated via equation, where \( f \) is the volume fraction of precipitates:

\[
\lambda = R_s \left( \sqrt{\frac{2\pi}{3f}} - \frac{\pi}{4} \right)
\]

The zirconium, ytterbium, and gadolinium content in the Al solid solution was 0.4, 0.2, and 0.2 wt.%, respectively. According to \([42]\), the equilibrium solubility of Zr, Yb, and Gd in a solid solution at 400 °C was 0.39, 0.1, and 0.1 wt.%. Except for the fact that Yb and Gd could dissolve in the \( Al_2Zr \) phase, the volume fraction of the phases could be estimated considering the density of simple \( L1_2 \) structured \( Al_3Zr, Al_3Yb, \) and \( Al_3Gd \) phases, which are \( 4.17 \) \([43]\), 5.69 \([42]\), and \( 4.98 \) \([44]\) g/cm\(^3\), respectively. The summarized volume fraction of the precipitates in the AlCuYbZr and AlCuGdZr alloys were estimated as 0.0054 and 0.0047, respectively. According to the TEM studies, the mean precipitate size was 50 ± 2 nm for the AlCuYbZr alloy and 42 ± 3 nm for the AlCuGdZr alloys (Figure 7). Following Equation 1, the contribution of the precipitates to YS was 32 MPa and 37 MPa for the AlCuYbZr and AlCuGdZr alloys, respectively. The volume fraction of the eutectic particles was 0.09 for both alloys. Considering particles with a size of 0.92, 1.04, 1.1, and 0.95 \( \mu m \) for AlCuYb, AlCuGd, AlCuYbZr, and AlCuGdZr alloys, their contributions to strength were 14, 12, 10, and 14 MPa, respectively. The solid solution strengthening effect of Cu in the Al matrix was calculated by the empirical equation \( \Delta \sigma_{SS} = 57.5 C_{\text{Cu}} \), where \( C_{\text{Cu}} \) is the atomic concentration of Cu \([45]\). The Cu content in the Al solid solution was 1.8 wt.% (0.77 at.%) and 1.6 wt.% (0.69 at.%) for the AlCuYbZr and AlCuGdZr alloys, respectively. Thus, the contribution of Cu atoms dissolved in the Al to the YS was 44 MPa for the AlCuYbZr alloy and 40 MPa for the AlCuGdZr alloy. On the other hand, Cu solute atoms provided the effect from the \( \theta'(Al_2Cu) \) phase after the aging treatment. Due to a low Cu content in (Al), the effect from the aging treatment was insignificant. For example, the maximum solubility of Cu in (Al) at 210 °C was 1.3%. The Cu content in (Al) in the investigated alloys was 1.6–1.8%. Volume fraction of \( \theta'(Al_2Cu) \) in this case was 0.0038–0.0057. The radius of \( \theta'(Al_2Cu) \) was about 75 nm. The unmodified Orovian calculation in this case demonstrated a low contribution from \( \theta'(Al_2Cu) \) of about 10–13 MPa.

Both alloys' sheets were processed in the same regimes, therefore we suggest a similar role for the dislocation and substructure strengthening mechanisms in the studied alloys. The similar structure was formed in the Zr-free alloys \([14]\), with the same content of Cu, Yb, and Gd. The eutectic originated particles and copper dissolved in the Al solid solution provided an equivalent contribution to the YS of the Zr-free alloys. Due to the nanoscale precipitates, the maximum strength values were observed in the AlCuYbZr and AlCuGdZr alloys sheets. YS values of 236–252 MPa and 225–237 MPa were achieved in the AlCuYbZr and AlCuGdZr alloy sheets after annealing at 210–250 °C, respectively. The YS of the AlCuYb and AlCuGd alloys sheets processed in the same regimes were 216–238 MPa and 175–202 MPa, respectively. These differences agreed with the calculated data and resulted from the contribution of nanoscale dispersoids to the strength.

AlCuYbZr and AlCuGdZr alloys are a prospective base composition for developing novel high technology heat resistance aluminum alloys. The strength of the alloy can be improved by alloying with Mn, Ti, and Mg, and their effect on the microstructure and mechanical properties needs to be studied.

4. Conclusions

The effect of zirconium on the microstructure, phase composition, and mechanical properties of the AlCuYb and AlCuGd alloys were studied in the present work. Zirconium dissolved completely in the aluminum solid solution of the AlCuYbZr and AlCuGdZr as-cast alloys. The AlCuYbZr/AlCuGdZr alloys demonstrated the same hardness vs. temperature and time relationships, but showed a higher hardness value compared to the AlCuYb/AlCuGd alloys due to precipitation of the \( Al_3(Zr,Yb) \) and \( Al_3(Zr,Gd) \) phases,
which were formed during the homogenization treatment and grew to 42–50 nm in diameter. The contribution of the precipitates to the YS was 33 MPa and 37 MPa for the AlCuYbZr and AlCuGdZr alloys, respectively.

The AlCuYbZr alloy had a higher yield and higher tensile strength than the AlCuGdZr alloy at the same annealing temperature and time due to a higher Cu content in the Al solid solution. The AlCuYbZr alloy exhibited good mechanical tensile properties at an annealing temperature of 100 °C for 1 h, with a yield strength of 276 MPa, ultimate tensile strength of 312 MPa, and elongation of 3.1%, while the as-rolled AlCuGdZr alloy had good mechanical tensile properties, with a yield strength of 279 MPa, ultimate tensile strength of 307 MPa, and elongation of 4.8%. The AlCuYbZr and AlCuGdZr alloys are a prospective base composition for developing novel high technology heat resistant aluminum alloys.

Author Contributions: Methodology and investigation, O.I.M.; data curation and methodology, S.M.A. and I.S.L.; visualization and formal analysis, M.V.G., A.G.M. and R.Y.B.; conceptualization, supervision, writing—review and editing, A.V.P. All authors have read and agreed to the published version of the manuscript.

Funding: The work was supported by the Russian Science Foundation (Project No. 21-79-00193).

Institutional Review Board Statement: Not applicable.

Informed Consent Statement: Not applicable.

Data Availability Statement: Not applicable.

Acknowledgments: The researcher S.M. Amer is funded by a scholarship under the joint (exclusive program between Arab republic of Egypt and Russia).

Conflicts of Interest: The authors declare no conflict of interest.

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