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Effect of Solution Heat Treatment Parameters on the Mechanical Properties of Re-aged EN AW-7022 Alloy

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The presented article deals with the influence of different solution annealing temperatures and times on the mechanical properties and microstructural changes of the EN AW-7022 aluminum alloy after two-stage re-aging. Following initial soft annealing, the specimens underwent solution treatment at 475 °C/ 20 min, 500 °C/20 min, 525 °C/30 min and 575 °C/30 min, followed by two-stage artificial aging at 120 °C/4 h and 175 °C/6 h. Mechanical properties were assessed using static tensile testing and Vickers hardness measurements. The highest strength levels were achieved after treatment at 525 °C/30 min. Metallographic observations together with SEM-EDS microanalysis confirmed a uniform distribution of strengthening precipitates of the Al_2CuMg and $Mg(Zn,Cu,Al)_2$ types at elevated temperatures, while fractographic examination indicated a predominantly ductile fracture mechanism. The results demonstrate that, even in the absence of additional plastic deformation, an advantageous combination of strength, hardness and ductility can be obtained through re-heat treatment. This is particularly beneficial for applications where dimensional stability and minimized residual stresses are required, such as in defense and aerospace components.

Keywords: Aluminium alloy, Heat treatment, Mechanical properties, Microstructure, Temperature

1 Introduction

Aluminum alloys are among the key materials in various technical fields, especially in the automotive, aviation and railway industries, where they are mainly used in the construction of lightweight and strength-demanding components [1–4]. They also find significant application in the field of defense industry and special equipment [5,6], where aluminum-based alloys are used for the production of pistons, crankcases, gearbox housings, engine blocks, cylinder heads, firearm components and armor [7–9]. The requirements placed on these materials significantly exceed the values of material characteristics of materials used in normal operation. Emphasis is placed on high strength, toughness, fatigue resistance, corrosion resistance. The advantages of aluminum alloys are low specific gravity, corrosion resistance, high thermal conductivity. The disadvantage of aluminum alloys lies in their lower mechanical properties, which are, however, necessary for their applications. Mechanical properties can be improved in several ways: alloying [10-12], coating [13-15] and especially heat treatment in the form of precipitation hardening [16-24].

Increased Zn content has been shown by Wen et al. [10] to increase the fatigue resistance of Al-Zn-Mg-Cu alloys. Among the shearable precipitates, the proportion of larger size precipitates for the higher zinc content alloy is bigger than that for the lower zinc content alloy. The coarse shearable precipitates hinder the propagation of the fatigue cracks, leading to inferior FCP (Fatigue Crack Propagation) rate. For both alloys, the shear mechanism possesses the dominant factor, finally causing a preponderance in the FCP resistance for the higher zinc content alloy than the lower one. The presence of Cu and Mg in the 7XXX series affects the kinetics and morphology of possible precipitates. Cu from the study [11] showed an increased effect of precipitation hardening through accelerated formation of GP zones during the pre-aging treatment and a finer distribution of T phase ($Mg_{32}(Al, Zn)_{49}$) in peak aged condition. However, Cu cannot significantly affect the kinetics of T phase formation, but on the other hand it increases its thermal stability. The higher the Mg content, the higher the volume fraction of formed precipitates. This can result in improved YS (Yield Strength) and decreased ductility [12].

The aging treatment of aluminum alloys is a key heat treatment process enhancing their strength and other mechanical properties, through the precipitation hardening mechanism, which is crucial for high-strength aluminum alloys such as 7XXX and 2XXX series alloys. The 7XXX series is used, among other things, in weapon manufacturing. These Al–Zn-Mg–Cu alloys are characterized by excellent mechanical properties and corrosion resistance. Several types of hardening are currently applied to achieve these properties: isothermal ageing, multi-stage ageing, non-isothermal ageing, retrogression and re-ageing (RRA), and stress ageing (i.e. creep ageing), which are the most widely used and most important of all.

The authors [1] studied the effect of ageing temperatures – UA (Under-aged), PA (Peak-aged), OA (Over-aged) on the mechanical properties of 7XXX alloys. The study of effects of ageing conditions of 7N01 and 7075 shows that the overaged samples have the lowest tensile strength among the three ageing states, but their fatigue strength is the highest. Since for the aging-strengthened Al alloys, different ageing states (such as under ageing and over ageing states) may contribute the same tensile strength. According to the selection principle of hardness, it is also found that the tensile strengths of the OA and UA states are similar for the experimental alloys.

In the article [4], the authors observed the effect of 2-stage aging on strength and corrosion resistance. They concluded that In two-stage aging of Al-Zn-Mg-Cu alloys, the corrosion resistance and mechanical strength compete with each other in the opposite direction. Therefore, it is necessary to optimize the involved aging parameters to obtain a desired combination of SCC resistance and mechanical response in 7xxx series Al alloys. However, there are very limited studies [7, 21] on the effects of solution temperatures on mechanical properties, especially on 7022 alloy [23]. This study aims to determine the variation of the mechanical behavior of EN AW-7022 alloy exposed to different solution temperatures and times.

The The main objective of this work is to analyze the influence of various solution annealing parameters—specifically temperature and holding time—on the mechanical properties of the EN AW-7022 aluminum alloy after re-aging. The study focuses on the evolution of hardness, tensile strength and fracture behavior in relation to different heat-treatment regimes, while the investigated material does not undergo any additional plastic deformation during processing. The aim is to identify heat-treatment parameters that provide mechanical properties as close as possible to the original condition, without the need for mechanical deformation. The material used in the experiment was supplied in the normalized T651 temper, which corresponds to solution annealing, artificial aging and subsequent stress relief by stretching (so-called stretching). The selection of solution heat-treatment temperatures was based on previously published data describing the dissolution behavior of strengthening phases in Al–Zn–Mg–Cu alloys. According to Jin et al. [25], temperatures around 475 °C represent the lower boundary at which η' ($MgZn_2$) and related phases begin to dissolve, although complete dissolution cannot always be achieved at this level. Experimental studies on AlZn5Mg3Cu alloys by Ganea-Christu et al. demonstrated that temperatures near 500 °C are commonly used to ensure sufficient dissolution of primary strengthening phases while maintaining microstructural stability [26]. Furthermore, research on Al–Mg–Si–Cu alloys by Liu et al. indicates that complete dissolution of more stable second-phase particles may require temperatures close to 550–560 °C, depending on alloy chemistry and heating conditions [30]. Based on these findings, solution-treatment temperatures of 475 °C, 500 °C, 525 °C and 575 °C were selected for this study [25–27].

2 Materials and methods

The experimental material used was the high-strength aluminum alloy EN AW-7022 ($AlZn_5Mg_3Cu$), which belongs to the group of Al-Zn-Mg-Cu alloys (7XXX series). The material was supplied in the form of rolled sheets with dimensions of 500 × 500 × 8 mm by BIKAR Metalle GmbH (Germany). This alloy is known for its high strengthening potential through precipitation hardening, as well as excellent strength, hardness and fatigue resistance, thanks to which it finds application mainly in the aerospace, armament and special industries. The chemical composition of the delivered material was compared with the data from the official material (certificate) and also experimentally verified using optical spectral analysis (OES). The results are shown in Table 1 and are expressed in weight percentages (wt %).

***Tab. 1** Chemical composition of EN AW-7022 (wt %)*

	Zn	Mg	Cu	Mn	Cr	Si	Fe	Ti+Zr	Al
Material sheet	4.3-5.2	2.6-3.7	0.5-1	0.1-0.4	0.1-0.3	max. 0.5	0.5	0.2	balance
Spectral analysis	4.46	3.45	0.61	0.17	0.13	0.07	0.01	0.04	balance

The material used in the experiment was delivered in the normalized condition T651, which includes solution annealing, artificial aging and subsequent removal of residual stresses by plastic deformation. The mechanical properties corresponding to this condition reached the values: tensile strength (Rm) 522 MPa, yield strength (Re) 480 MPa, elongation (A) 11% and Vickers hardness (HV1) 181. Test specimens intended for static tensile testing were manufactured from the delivered sheet using a DATRON M8Cube high-speed CNC machining center (Fig. 1). All manufactured samples were subsequently subjected to soft annealing at a temperature of 400 °C for 30 minutes. The heat treatments were carried out in a Nabertherm LHT laboratory furnace with a certified temperature accuracy of ±2 °C. Part of the thus annealed samples was further processed by precipitation hardening, which consisted of solution annealing at different temperatures and times, followed by two-stage aging (Table 2). Five samples were produced for each heat treatment variant in order to ensure statistical reliability of the measurements.

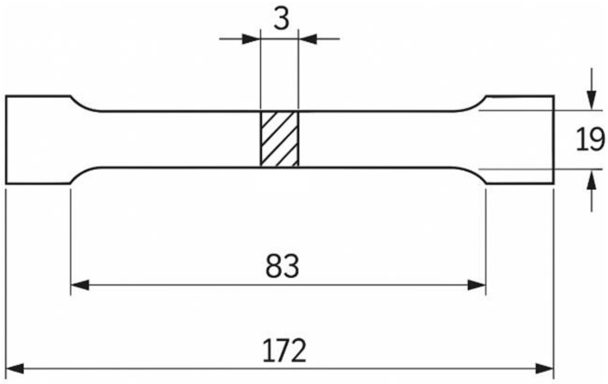

***Fig. 1** Shape and dimensions of test specimens for static tensile testing*

***Tab. 2** Hardening of alloy EN AW-7022*

Designation	Solution annealing		Cooling	Ageing	
	T [°C]	t [min]		Stage [°C/hr]	Stage [°C/hr]
1	475	30			
2	525	30			
3	575	30	water	120/24	170/6
4	475	20			
5	500	20			

The static tensile test was performed on an Instron 5500R universal testing machine in accordance with the requirements of the STN EN ISO 6892-1 standard, which sets out a method for testing metallic materials at room temperature. The Vickers hardness was determined on a Qness 250 CS EVO device in accordance with the EN ISO 6507 standard, with several measurements being made for each sample in order to obtain an average value. After the mechanical tests, metallographic analysis was performed on all samples, which included the preparation of metallographic sections and subsequent etching of the microstructure. Etching was performed using Keller's solution, which is standardly used for 7XXX series aluminum alloys. This etching solution consisted of the following chemical composition: 95 ml distilled water, 2.5 ml HNO_3, 65%, 1.5 ml HCl, 37% and 1 ml HF, 40%. The solution used allowed for the highlighting of structural components such as grain boundaries, segregations and precipitation phases, thus creating suitable conditions for microscopic evaluation using optical microscopy.

3 Results and discussions

3.1 Local microchemical analysis by SEM-EDS

Fig. 2 shows the results of the micro-chemical analysis of the EN AW-7022 alloy performed using scanning electron microscopy with energy-dispersive X-ray spectroscopy (SEM-EDS). Since SEM-EDS provides only local, semi-quantitative information, the obtained results were used exclusively to characterise the distribution of alloying elements and the composition of secondary phases, not to determine the bulk chemical composition of the material. The elemental map in Fig. 2a confirms a uniform distribution of aluminium as the primary matrix element. Local enrichments of Zn, Mg and Cu were detected in several micro-regions. These are typical for the presence of metastable strengthening precipitates, especially η' ($MgZn_2$) and S (Al_2CuMg) phases, which commonly form in Al–Zn–Mg–Cu alloys after ageing treatments. The detected presence of trace amounts of Fe, Mn, Cr and Ti is consistent with their function as grain stabilizers or technological additives, which in

7xxx alloys typically appear in quantities below ~0.2 wt.% and contribute to structural refinement and improved thermal stability. The representative point spectrum in Fig. 2b demonstrates the semi-quantitative elemental proportions in one such precipitate-rich locality. The spectrum shows the dominant presence of aluminium (~80.3 wt.%), accompanied by locally increased concentrations of zinc (~6.5 wt.%), magnesium (~3.6 wt.%), and copper (~1.1 wt.%). These values are characteristic for micro-areas containing Zn- and Mg-rich precipitates, and therefore do not represent the global alloy composition. Because EDS cannot reliably determine the overall chemical composition of the alloy due to its limited penetration depth and surface sensitivity, the bulk composition was subsequently determined using OES. The OES results confirmed that the alloy fully complies with the standard composition limits for EN AW-7022. This verifies that the material used in the experiment represents a standard AlZn5Mg3Cu alloy suitable for the subsequent heat-treatment analysis. Overall, the SEM-EDS observations provide complementary micro-chemical information that supports the interpretation of precipitation processes and microstructural evolution discussed in later sections.

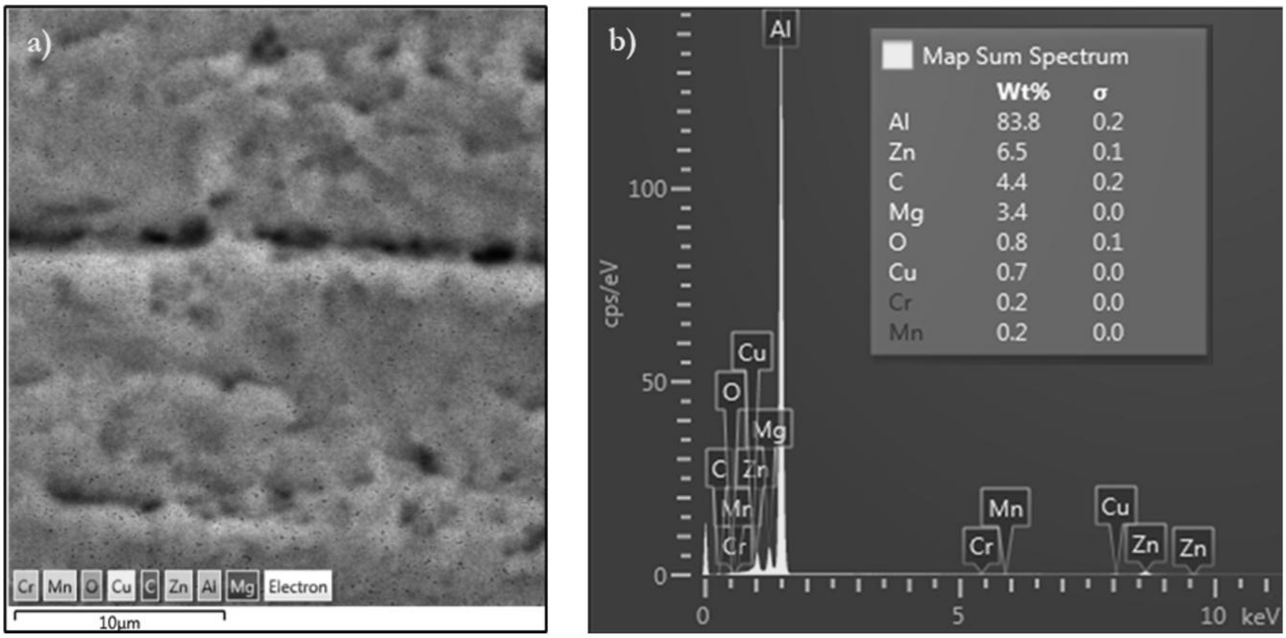

Fig. 2 *a) Spectral map of selected elements; b) EDS spectrum of the chemical composition of the EN AW-7022 alloy analyzed by SEM-EDS*

3.2 Analysis of changes in mechanical properties

The results of the static tensile test - Rm, Re, A and HV1 hardness are shown in Table 3 and Fig. 3. The samples subjected to soft annealing and again 2-stage aging show lower values of tensile strength than the samples in the delivered state T651. The decrease is from 4 to 53%. The most significant decrease was recorded in sample 4, where the lowest temperature of solution annealing and the shortest time (475 °C/20min) were used. We assume that such a significant decrease compared to the delivered state T651 was caused by insufficient dissolution of the $Mg(Zn,Cu,Al)_2$ and Al_2CuMg phases. The same is observed in sample 5, where the temperature of 500 °C was applied but only for 20 min. This was manifested by a decrease in tensile strength by 40.5% compared to the delivered state T651.

In the case of using higher solution annealing temperatures 525 °C (sample 2) and 575 °C (sample 3) for 30min, we observed a decrease in strength compared to T651 by 4% for sample 2 and by 29% for sample 3. When applying repeated hardening with two-stage aging, the use of the solution annealing regime of 525°C/30min was demonstrably confirmed. Zn and Mg elements are highly soluble in the matrix of 7xxx alloys. During the ageing treatment, these elements are forced to push out from supersaturated crystal lattice due to the decreased matrix solubility. This is the reason, why the atoms are aggregated into small precipitates (secondary phase). This secondary phase precipitation increases the strength of the material by inhibiting dislocation movements [20]. The recommended solution annealing temperature is 470 – 480 °C in the material data sheet. A more pronounced downward trend was observed for the yield strength, where it represented a decrease in the range of 22 – 80%. The most pronounced was for sample 4, where the yield strength dropped by up to 80%. The smallest decrease was shown by sample 2, namely 22%. The aim of hardening is to increase the strength characteristics (Rm, Re) at the expense of the plastic characteristics (A). However, during repeated hardening, we observe an increase in ductility compared to the delivered state – samples 1, 2, 4, 5. A decrease is observed for sample 3, namely in the value of 45%. The increase in ductility is recorded in the range of 145 – 164%. The trend of increasing ductility is not identical to the trend of decreasing Rm and Re. Samples 1, 2, 4, 5, despite different temperatures and times of solution annealing, showed very close ductility values - 16 - 18%, which does not correspond to the influence of temperature on Rm and Re. The exception is sample 3, which was the only one to show a 55% decrease in ductility.

Tab. 3 *Mechanical properties of EN AW-7022*

Sample	Rm [MPa]	Re [MPa]	A [%]	HV1
Z	522 ± 5	480 ± 5	11± 1	181 ± 5
1	431 ± 4	241 ± 3	16 ± 1	122 ± 4
2	499 ± 4	373 ± 4	17 ± 1	151 ± 3
3	370 ± 3	307 ± 4	6 ± 1	174 ± 4
4	244± 2	307 ± 4	18 ± 2	65 ± 2
5	269 ± 2	96 ± 2	17 ± 2	68 ± 2

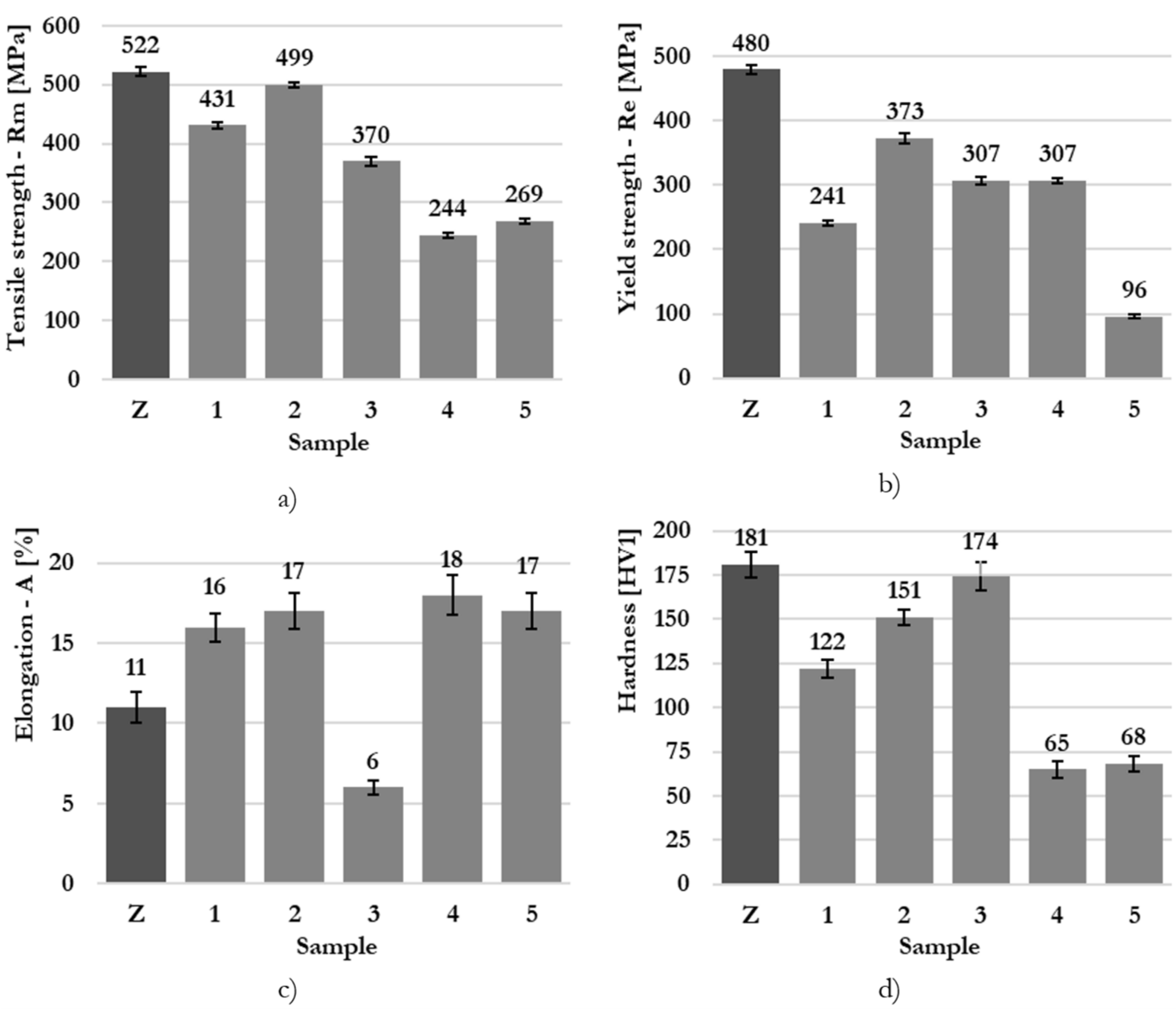

Fig. 3 *Mechanical properties of EN AW-7022: a) Tensile strength b) Yield strength c) Elongation d) Hardness*

The Vickers hardness is affected by the temperature of solution annealing, which was reflected in different measured values, which all recorded a decrease compared to the delivered state in the range of 64 - 3%. The most significant decrease was observed in samples 4 and 5 - higher temperatures of solution annealing and shorter times. Samples 1, 2, 3 showed a decrease proportional to the values of tensile strength and yield strength. The HV1 hardness values at higher solution annealing temperatures (525 and 575 °C) recorded a smaller decrease than at lower temperatures (475 and 500 °C). However, even at higher solution annealing temperatures, the hardness values as in the as-delivered T651 condition were not achieved. The maximum hardness value achieved was for sample 3 – 174 HV1. The main reason for the increase in hardness lies in the formation of a secondary phase of precipitates in the matrix as a result of hardening. These precipitates prevent the movement of dislocations [20].

3.3 Microstructure analysis

From the thermodynamic equilibrium it follows that although the chemical composition, characteristic secondary phases and hardening parameters of individual wrought Al alloys are different, the precipitation process of the secondary phase and the transformation of the deformation mechanism have many similarities. The precipitation process of the second phases includes the nucleation and precipitation of GP (Guinier–Preston) zones, the transformation from GP zones to a metastable phase, and the growth and coarsening of the metastable phase into an equilibrium phase. The finely dispersed GP zones and metastable phases are all coherent or semi-coherent with the matrix. These similarities in the micro-scopic mechanisms should be reflected also in the mechanical properties of wrought Al alloys [19]. The purpose of heat treatment for Al 7XXX alloys is to optimize the three microstructure parameters of matrix precipitates (MP), grain boundary precipitates (GBPs), and precipitate-free zone (PFZ), so that the alloys have good comprehensive properties [7]. It is generally accepted that the width of the precipitation-free zone (PFZ) is correlated with the alloys strength; and narrower PFZ leads to higher strength [11]. Solid solution is the basis of the heat-treated strengthening aluminum alloys to obtain high strength. It aims to fully dissolve the soluble elements in the alloy into the aluminum matrix to form a nearly uniformly distributed supersaturated solution, which facilitates subsequent aging precipitation to strengthen the alloy. The basic operations of the solution treatment are heating and holding. The solution temperature and holding time are the two most important parameters that determine the effect of the solution. If the solution temperature is higher and the holding time is longer, the diffusion of solute atoms is more favorable, so that the alloy elements are more fully dissolved and the aging effect is better [7].

The microstructure of the hardened aluminum alloy 7022 (Fig. 4a) is formed by a solid solution α – Al(FeCrSi) and secondary phases $Mg(Zn,Cu,Al)_2$ – T phase and Al_2CuMg – S phase [28]. The delivered material is formed by uniformly distributed precipitates of the T and S phases in the matrix of the solid solution α. Fine precipitates with high density reduced the PFZ and significantly increased the strength, which reaches 522 MPa. A very similar distribution of precipitates, without their exclusion at the grain boundaries, was observed in sample 2 (Fig. 4c). The distribution of precipitates showed a linear character, which was also reflected in a higher Rm value compared to samples 1, 3, 4 and 5, but did not reach the value of the sample in the delivered state. Precipitates excluded in the matrix with signs of linearity were also observed in sample 3 (Fig. 4d). The precipitates were of different sizes, which caused a larger PFZ and thus a lower strength (370 MPa). For samples 2 and 3, a higher solution annealing temperature (525 °C and 575 °C) was used, which ensured complete dissolution of the S phase and diffusion of Cu [18]. The application of lower solution annealing temperatures (475 °C, 500 °C), or. at a temperature of 500 °C for a shorter time – 20 min, caused incomplete dissolution of the S phase and insufficient diffusion of Cu, which is significantly slower than the diffusion of Mg and Zn. This resulted in the formation of not only fine dispersed precipitates of the secondary phase but also a discontinuous network of the secondary phase at the grain boundaries after two-stage aging (Fig. 4 b, e, f). The consequence is a significantly lower strength depending on the amount of the discontinuous network. The smallest amount was observed in sample 1 and we achieved a strength of 431 MPa. For samples 4 and 5 the amount was almost identical and the strength values reached 244 MPa (sample 4) and 269 MPa (sample 5).

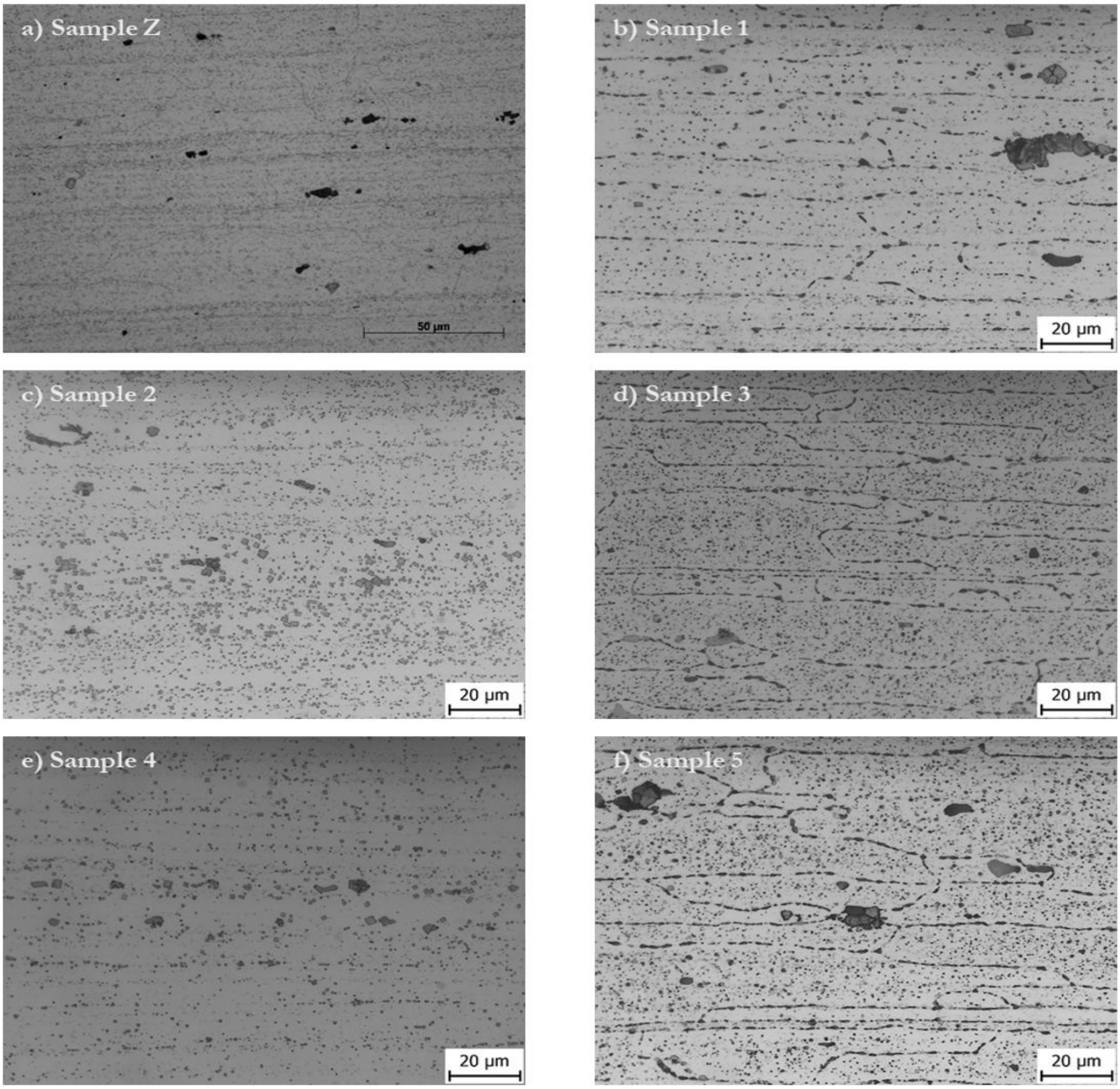

Fig. 4 *Microstructure of EN AW-7022 after different solution temperatures: a) T651, b) 475 °C/30min, c) 525 °C/30min, d) 575 °C/30min, e) 475 °C/20min, f) 500 °C/20min*

3.4 Fractographic analysis

Fractographic analysis of fracture surfaces of individual samples of the EN AW-7022 alloy (Fig. 5) revealed a homogeneous nature of material failure, with the dominant mechanism being microvoid coalescence fracture, typical of ductile behavior of metallic materials. The fracture morphology was characterized in all cases by pronounced dimples (ductile dimples), which arise as a result of nucleation, growth and subsequent coalescence of micropores in the matrix during plastic deformation before fracture. A more detailed examination of the fracture surfaces did not identify signs of brittle failure, such as intercrystalline cracks or split areas, which indicates sufficient toughness of the material even after repeated heat treatment without further plastic deformation. All samples showed a uniform nature of failure, without the presence of significant heterogeneities, which correlates with the results of EDS and metallographic analyses, where the relative homogeneity of the distribution of alloying elements was confirmed. The presence of alloying elements such as Zn, Mg and Cu has a significant impact on the mechanical properties and fractography of the material. These elements form precipitation phases during the aging process – especially $Mg(Zn,Cu,Al)_2$ and Al_2CuMg – which increase the strength of the alloy by inhibiting the movement of dislocations. Their uniform dispersion contributes to the uniform distribution of stresses in the microstructure, thereby reducing the risk of localized failure. Based on the observed fractographic features, it can be concluded that the thermal regime applied to the alloy did not affect the fracture mode and the material retained the desired plasticity and fracture toughness, which is favorable for applications requiring reliability under static and dynamic stress.

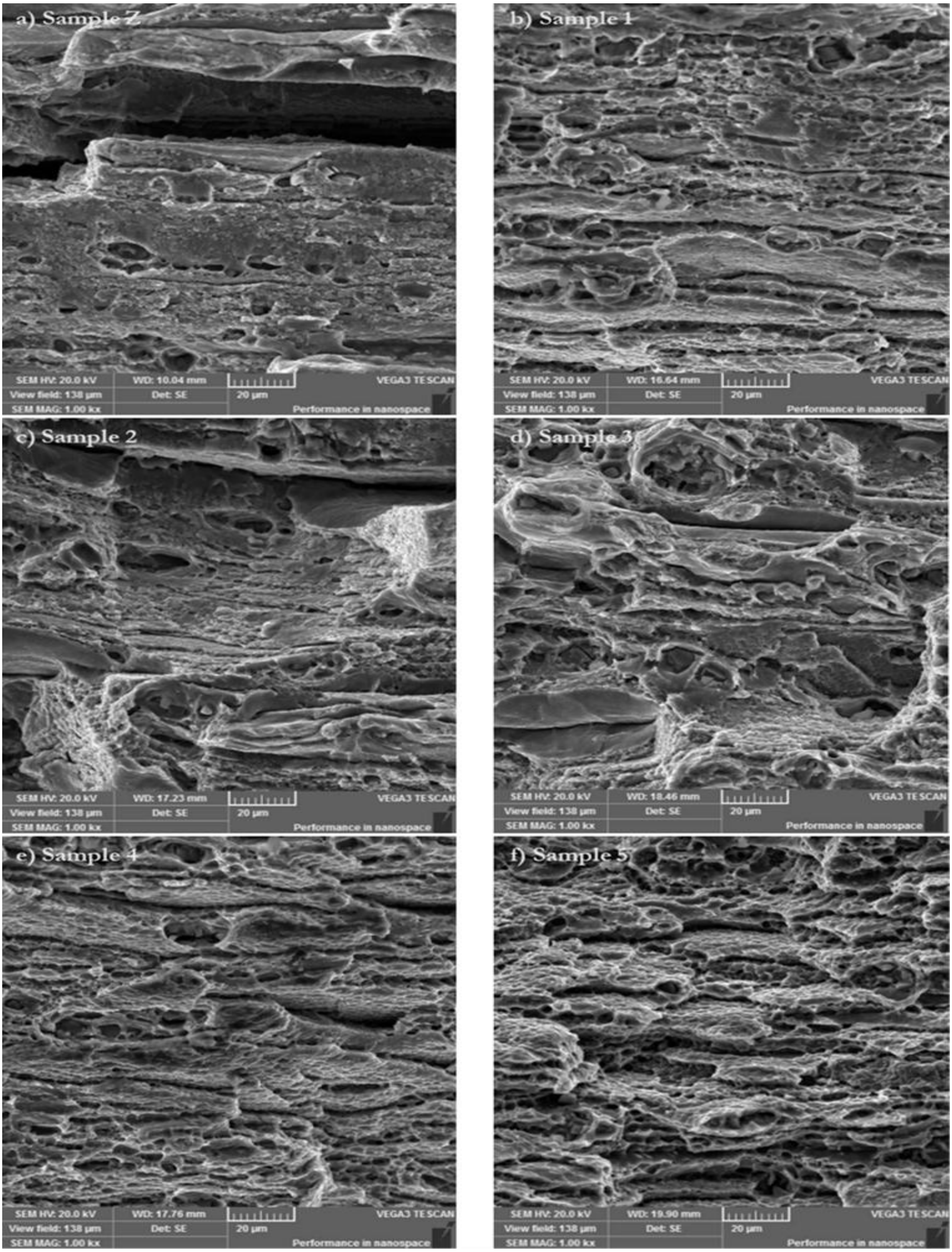

***Fig. 5** Fractographic analysis of fracture surfaces of EN AW-7022 samples after tensile test: a) T651, b) 475 °C/30min, c) 525 °C/30min, d) 575 °C/30min, e) 475 °C/20min, f) 500 °C/20min*

4 Discussion

The strength characteristics (Rm, Re) after repeated heat treatment decrease by 4–53% compared to the T651 state, with the most significant decrease being recorded when applying lower solution annealing temperatures (475 °C/20min and 500 °C/20min). This phenomenon is closely related to the insufficient dissolution of strengthening precipitates of the $Mg(Zn,Cu,Al)_2$ and Al_2CuMg types, the complete dissolution of which is a necessary prerequisite for effective precipitation strengthening during subsequent artificial aging. He Yin, Zhi Hui Li et al. confirmed in their study that almost complete dissolution of $Mg(Zn,Cu,Al)_2$ phases occurs already at 465 °C/2h and that higher temperatures and longer times are required for Al_2CuMg to be effectively eliminated from the matrix [29]. Similarly, Li et al. in their study emphasize that insufficient dissolution of phases after solution annealing limits the potential of strengthening mechanisms, which is manifested by reduced hardness and increased plasticity – which is also in line with our observations [30]. Also, the statement by Zhang, X et. al that: "Residual precipitates that are not completely dissolved reduce the hardening potential during subsequent natural and artificial aging", directly stops the influence of residual precipitates on the resulting mechanical properties [27]. On the contrary, the use of temperatures of 525 °C and 575 °C led to a smaller decrease in strength, with the most favorable properties being recorded at the regime of 525 °C/30min. The higher temperature of 575 °C probably led to partial overheating and grain growth, which could have reduced the efficiency of precipitation strengthening. In addition to precipitate evolution, repeated heat treatment may also contribute to grain coarsening, which can influence the resulting mechanical properties. According to the Hall–Petch relationship, an increase in grain size reduces the resistance to dislocation motion and consequently decreases the yield strength and tensile strength. Therefore, prolonged exposure to elevated temperatures during repeated solution treatment may contribute not only to precipitate coarsening and overaging, but also to a reduction in grain-boundary strengthening. As reported by Liu et al., too long exposure at high temperatures can lead to coarsening of precipitates, which results in a decrease in the barrier effect against dislocation movement [31]. The same trend is also confirmed by the study of Zou et al., where a decrease in strength was observed with excessive growth of η' phase precipitates, which directly reduces the lattice reinforcement effect and worsens the thermal stability of the material [32]. In terms of ductility (A), an improvement in plastic properties was surprisingly observed in most of the rehardened samples. This phenomenon may be attributed to the reduced dislocation density due to the absence of plastic deformation and larger precipitate sizes, which are less hindering for dislocation movement, as was also observed in the work of Ren, X et al., where the authors describe the influence of precipitate evolution on the balance between strength and ductility in Al-Zn-Mg-Cu alloys [33]. The measured hardness values correlated with the results of strength tests – higher hardness values were recorded at higher solution annealing temperatures, which indicates a more efficient formation of precipitates during aging. Nevertheless, even under optimal conditions, the original hardness values of the T651 state (181 HV1) were not achieved. This decrease can be attributed to the absence of plastic deformation, which in the production process contributes to an increase in the density of dislocations and thus to a more efficient precipitation hardening during aging [34]. From a practical point of view, however, the absence of mechanical deformation is an advantage – it preserves the exact dimensions of the components, prevents the formation of residual stresses or deformations and optimizes the production process by effectively reducing technological operations, which is especially desirable in the aerospace and armament industries [35, 36].

5 Conclusion

The presented article deals with the influence of solution annealing temperature on the mechanical properties of EN AW-7022 aluminum alloy after previous soft annealing. The aim was to eliminate the influence of previous treatment in the T651 condition and to assess the potential of repeated two-stage hardening at different solution annealing parameters. Based on experimental tests (static tensile test, Vickers hardness measurement), metallographic observations and fractographic analysis, the following conclusions were formulated:

- The EN AW-7022 alloy did not achieve higher strength or hardness after repeated two-stage heat treatment compared to the original T651 condition. The decrease in strength properties (Rm, Re) ranged from 4–53%, most significantly at low solution annealing temperatures (475 °C/20 min and 500 °C/20 min), which is related to insufficient dissolution of strengthening precipitates of the $Mg(Zn,Cu,Al)_2$ and Al_2CuMg type.
- The highest tensile strength was achieved at a temperature of 525 °C/30 min, with this

sample 2 reaching values of Rm = 499 MPa and Re = 373 MPa. This regime ensured an optimal balance between phase dissolution, grain growth and precipitation strengthening efficiency.

- The highest hardness (174 HV1) was recorded for the sample processed at 575 °C/30 min, which indicates a higher volume of secondary precipitates. However, even this regime did not reach the original hardness of the T651 state (HV1 = 181), which can be attributed to the absence of plastic deformation before hardening.
- An increase in ductility (A) was observed in most of the re-hardened samples 1, 2, 4, 5. This phenomenon is related to the lower dislocation density and larger precipitate sizes, which are less restrictive of dislocation movement.
- The most suitable heat treatment regime was 525 °C/30 min, which ensured the optimal combination of mechanical properties without signs of overheating. At the same time, this temperature allowed for the effective dissolution of Al_2CuMg-type phases and the uniform exclusion of secondary precipitates during aging, which contributed to the homogeneous microstructure and stable material properties.
- Fractographic analysis of all examined samples revealed a dominantly ductile failure mechanism. The pitted fracture structure confirmed microvoid coalescence as the main failure mechanism, without the presence of brittle fractures or grain separations. Typical plastic destruction indicates the stability of the matrix phase and the uniform distribution of precipitates.
- From a technological point of view, rehardening without the application of plastic deformation is advantageous especially in the production of precision parts, where it is necessary to eliminate residual stresses, reduce the number of manufacturing operations and ensure dimensional stability of components - which is extremely desirable especially in the aerospace and armaments industries.

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