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CAST ALUMINIUM ALLOYS AS COMPETITIVE MODERN STRUCTURAL MATERIALS: CHEMICAL COMPOSITIONS, FABRICATION, TREATMENT, TESTING, MECHANICAL PROPERTIES, AND APPLICATIONS

The paper presents an overview of modern aluminium alloys, the technologies used to produce and process them, and the formation of their microstructure and properties. Currently, mechanical engineering has a trend toward the widespread use of resource-

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efficient, environmentally friendly, and high-precision technologies. Since aluminium alloys have low density, relatively high strength, sufficient ductility, and certain special properties (rather high strength at elevated temperatures), they are ideal materials for widespread use in the automotive, instrumentation, and aerospace industries. A classification of aluminium alloys with different groups is provided, based on the alloying elements. Analogies are drawn between the national (Ukrainian) classification of the alloys and those used in leading industrialised countries. The similarities and differences in the chemical compositions and properties of the alloys are analysed. Casting and additive manufacturing techniques for producing aluminium-alloy parts are considered. The effect of structure-forming processes during the solidification of the alloys on their phase composition and mechanical properties is shown. Various alloying elements are considered, and their effect on the microstructure and properties is analysed. Content limits for each alloying element and the main technological features of alloy melting are recommended. Alloy modification techniques are also discussed. The effect of various modifying elements on the morphology of the overall structure and the hardening phases is shown. A separate section of the paper provides an overview of various heat treatments for cast aluminium alloys. These alloys are highly sensitive to thermal effects. Therefore, selecting the appropriate heat-treatment conditions is key for a significant enhancement of the strength (sometimes, by a factor of 1.5–2.0) through the formation of additional structural (phase) constituents. In particular, machining procedures are determined for parts used under extreme conditions (high temperatures of above 300 °C, corrosive environments, *etc*., as in the aviation or aerospace industries). The most common intermetallic phases in the aluminium alloys are listed; the effect of each of these phases on the properties is discussed. A comparative diagram is built, which shows the mechanical properties (strength and elongation) of the alloys commonly used and those ones under development (experimental alloys). The data shown in this diagram combine the results of many years of studies conducted by researchers from various countries, which were published in leading scientific journals. The data on the strength at elevated temperatures are also presented; the mechanical properties at high temperatures (up to 450 °C) and upon prolonged exposure to extreme conditions are considered. The current scientific and technological issues, on which the experts in aluminium casting are currently working, are identified.

Keywords: aluminium-casting alloys, alloying, heat treatment, mechanical properties, structure formation, recycling.

1. Classification of Cast Aluminium Alloys and Their Chemical Composition

The industrial production of aluminium-based alloys is currently undergoing rapid development. This leads to the development of new (experimental) alloys and innovative approaches to improving the chemical composition of conventional alloys. For this reason, some alloys that were previously widely applied in the industry (such as AK12pch (12% Si, high purity), AMg11 (11% Mg), AM4.5Cd (up to 0.25% Cd), and several others) are no longer used for the manufacture of cast parts. They have been replaced by new alloys with superior mechanical and special properties, which are considered in this review.

The integration of the Ukrainian foundry industry into the European and global market involves a gradual transition from state (domestic) standards and regulations to those that are universally recognised in global practice. As a result, the former Soviet GOST 1583-93 and DSTU 2839-94 (an exact copy of that GOST) were initially replaced by DSTU EN 1706:2006, and then DSTU EN 1706:2022 'Aluminium and aluminium alloys. Casting. Chemical composition and mechanical properties'.

Under the previous standards, alloys were classified into five separate groups [1]. There were 39 grades of cast aluminium alloys:

I — 11 alloys of the Al–Si–Mg system;
II — 14 alloys of the Al–Si–Cu system;
III — 2 alloys of the Al–Cu system;
IV — 9 alloys of the Al–Mg system;
V — 3 Al-based alloys with other elements.

According to the current DSTU EN 1706:2022 standard, which is fully compliant with international standards, there are 45 cast aluminium alloys. There is no distinct division of these alloys into groups. Therefore, a general overview of the chemical compositions of these alloys, as well as those that are not included in the national standard but are globally used, is provided.

The alloys of the Al–Si system with their chemical compositions are listed in Table 1.

The alloys of the Al–Si–Cu system with their chemical compositions are listed in Table 2.

The alloys of the Al–Cu system with their chemical compositions are listed in Table 3. The equivalents of the alloys in this group are the 2000 series alloys, which contain additional alloying elements and are listed in the Table.

Table 1. **Alloys of the Al–Si system and their chemical compositions**

#	Alloy	Si	Cu	Mn	Fe	Impurities (Mg, Cr, Ni, Zn, Pb, Sn, Ti, *etc.*), no more than
1	$AlSi_{11}$ (AC-44000)	10.0–11.8	≤0.05	≤0.10	≤0.19	0.77
2	$AlSi_{12}$(a) (AC-44200)	10.5–13.5	≤0.05	≤0.35	≤0.55	0.40
3	$AlSi_{12}$(b) (AC-44100)	10.5–13.5	≤0.15	≤0.55	≤0.65	0.80
4	$AlSi_{12}$(Fe)(a) (AC-44300)	10.5–13.5	≤0.10	≤0.55	0.45–0.90	0.55
5	$AlSi_{12}$(Fe)(b) (AC-44100)	10.5–13.5	≤0.20	≤0.55	0.45–0.90	1.10
6	$AlSi_9$ (AC-44400)	8.0–11.0	≤0.10	≤0.50	≤0.65	0.70
7	$AlSi_{10}Mn$ (AC-44600)	9.5–11.5	≤0.03	0.30–0.75	≤0.25	0.53
8	$AlSi_{12}$(Cu) (AC-47000)	10.5–13.5	0.9–1.0	≤0.55	≤0.80	2.00
9	$AlSi_{12}Cu_1Fe$ (AC-47100)	10.5–13.5	0.7–1.2	≤0.55	0.60–1.10	2.00
10	$AlSi_{12}$(Fe) (AC-47200)	10.5–13.5	≤0.40	≤0.50	0.60–1.10	1.85

Table 2. **Alloys of the Al–Si–Cu system and their chemical compositions**

#	Alloy	Content, wt.%					
		Si	Cu	Fe	Mg	Others	Impurities (Mn, Cr, Zn, Ni, Pb, Sn, Ti, etc.), no more than
1	$AlSi_6Cu_4$ (AC-45000)	5.0–7.0	3.0–5.0	≤1.0	≤0.55	—	3.95
2	$AlSi_5Cu_3Mg$ (AC-45100)	4.5–6.0	2.6–3.6	≤0.60	0.15–0.45	—	1.40
3	$AlSi_5Cu_1Mg$ (AC-45300)	4.5–5.5	1.0–1.5	≤0.65	0.35–0.65	—	1.55
4	$AlSi_5Cu_3$ (AC-45400)	4.5–6.0	2.6–3.6	≤0.60	≤0.05	—	1.40
5	$AlSi_7Cu_{0.5}Mg$ (AC-45500)	6.5–7.5	0.2–0.7	≤0.25	0.20–0.45	—	0.52
6	$AlSi_7Cu_1Mg_{0.6}$ (AC-45600)	6.5–7.5	0.8–1.6	≤0.19	0.45–0.70	—	0.52
7	$AlSi_9Cu_3$(Fe) (AC-46000)	8.0–11.0	2.0–4.0	0.6–1.1	≤0.55	—	3.40
8	$AlSi_{11}Cu_2$(Fe) (AC-46100)	10.0–12.0	1.5–2.5	0.45–1.0	≤0.30	—	3.75
9	$AlSi_8Cu_3$ (AC-46200)	7.5–9.5	2.0–3.5	≤0.80	≤0.55	—	3.10
10	$AlSi_7Cu_3Mg$ (AC-46300)	6.5–8.0	3.0–4.0	≤0.80	0.30–0.60	—	2.35
11	$AlSi_9Cu_1Mg$ (AC-46400)	8.3–9.7	0.8–1.3	≤0.80	0.25–0.65	—	2.20
12	$AlSi_9Cu_3FeZn$ (AC-46500)	8.0–11.0	2.0–4.0	0.6–1.2	≤0.55	Zn 3.0	2.20
13	$AlSi_7Cu_2$ (AC-46600)	6.0–8.0	1.5–2.5	≤0.80	≤0.35	—	2.80
14	$AlSi_{12}CuMgNi$ (AC-48000)	10.5–13.5	0.8–1.5	≤0.70	0.8–1.5	Ni 0.7–1.3	1.10
15	$AlSi_{17}Cu_4Mg$ (AC-48100)	16.0–18.0	4.0–5.0	≤1.30	0.25–0.65	—	2.70
16	$AlSi_{15}Cu_3MgFe$ (AC-48200)	14.5–16.5	3.0–4.0	0.7–1.2	0.55–0.90	—	2.90

The alloys of the Al–Mg system with their chemical compositions are listed in Table 4.

The alloys of the Al–Si–Mg system with their chemical compositions are listed in Table 5.

The alloys of the Al–Zn system with their chemical compositions are listed in Table 6 (in DSTU EN standards, only one alloy is specified). Their closest equivalents are the 7000 series alloys, which are also shown in the Table.

Al–Li alloys constitute a separate group and are included in the 2000 series. The chemical compositions of these alloys are given in Table 7.

Table 3. **Alloys of the Al–Cu system and their chemical compositions**

#	Alloy	Content, wt.%						
		Cu	Mn	Ti	Mg	Zr	Others	Impurities (Si, Fe, Ni, Cr, Pb, Sn), no more than
1	$AlCu_4MgTi$ (AC-21000)	4.2–5.0	≤0.10	0.15–0.30	0.15–0.35	—	—	0.90
2	$AlCu_4Ti$ (AC-21100)	4.2–5.2	≤0.55	0.15–0.3	–	—	—	0.55
3	201.0, CO51A, 382	4.6	≤0.35	0.15–0.35	≤0.35	—	Pb 0.7 Ag 0.7	Fe 0.15 Si 0.10
4	AC1B	4.0–5.0	≤0.10	0.05–0.30	0.15–0.35	—	—	0.70
5	A-U5GT	4.2–5.0	≤0.10	0.05–0.30	0.15–0.35	—	—	Si 0.70
6	2024 [4]	3.7–4.4	0.5–0.6	0.02–0.15	1.2–1.5	—	—	1.10
7	2026 [4]	3.6–4.3	0.3–0.8	≤0.06	1.0–1.6	0.05–0.25	—	Fe 0.07 Si 0.05
8	2219 [5, 6, 7]	6.3–6.5	0.3–0.32	0.03–0.06	—	0.15	V 0.06	Fe 0.13 Si 0.05
9	2224 [4]	4.1	0.6	—	1.5	—	—	Fe 0.15 Si 0.12
10	2324 [4]	3.8–4.4	0.3–0.9	0.15	1.2–1.8	—	—	Cr 0.10 Fe 0.12 Si 0.10
11	2524 [4]	4.0–4.5	0.45–0.7	≤0.10	1.2–1.6	—	—	Cr 0.05 Fe 0.12 Si 0.06
12	206 [8]	4.5	≤0.23	≤0.02	≤0.30	—	—	Fe 0.14 Si 0.12
13	Al_7Cu [8]	6.4	≤0.19	≤0.09	—	0.13	—	Fe 0.10 Si 0.05
14	RR350 [8]	4.8	≤0.19	0.21	—	0.17	—	Fe 0.09 Si 0.05

Table 4. **Alloys of the Al–Mg system and their chemical compositions**

#	Alloy	Mg	Si	Mn	Impurities (Fe, Cu, Cr, Ni, Zn, Pb, Sn, Ti, *etc.*), no more than
1	$AlMg_3$ (AC-51100)	2.5–3.5	≤0.55	≤0.45	1.05
2	$AlMg_9$ (AC-51200)	8.0–10.5	2.5	≤0.55	2.00
3	$AlMg_5$ (AC-51300)	4.3–6.8	≤0.55	≤0.45	1.10
4	$AlMg_5$(Si) (AC-51400)	4.5–6.5	1.5	≤0.45	1.05
5	$AlMg_5Si_2Mn$ (AC-51500)	4.7–6.0	1.8–2.6	0.40–0.80	0.80

Table 5. **Alloys of Al–Si–Mg system and their chemical compositions**

Content, wt.%							
#	Alloy	Si	Mg	Mn	Ti	Others	Impurities (Fe, Cu, Cr, Ni, Zn, Pb, Sn, *etc.*), no more than
1	$AlSi_2MgTi$ (AC-41000)	1.6–2.4	0.45–0.65	≤0.50	0.05–0.2	—	1.10
2	$AlSi_7Mg$ (AC-42000)	6.5–7.5	0.25–0.65	≤0.35	≤0.25	—	1.40
3	$AlSi_7Mg_{0.3}$ (AC-42100)	6.5–7.5	0.25–0.45	≤0.10	≤0.25	—	0.40
4	A357 Al-$Si_7Mg_{0.6}$ (AC-42200)	6.5–7.5	0.45–0.70	≤0.10	≤0.25	—	0.40
5	$AlSi_7$(Mg) (AC-42300)	6.5–7.5	0.10–0.25	≤0.10	≤0.25	—	0.40
6	$AlSi_7MnMg$ (AC-42400)	6.5–8.5	0.10–0.45	0.35–0.75	≤0.20	—	0.40
7	$AlSi_{10}Mg$ (AC-43000)	9.0–11.0	0.25–0.45	≤0.45	≤0.15	—	1.00
8	$AlSi_{10}Mg$(Cu) (AC-43200)	9.0–11.0	0.20–0.45	≤0.55	≤0.20	Cu 0.35	1.40
9	$AlSi_9Mg$ (AC-43300)	9.0–10.0	0.20–0.45	≤0.10	≤0.15	—	0.40
10	$AlSi_{10}Mg$(Fe) (AC-44000)	9.0–11.0	0.20–0.50	≤0.55	≤0.20	Fe 0.45–0.9	0.75
11	$AlSi_{10}MnMg$ (AC-43500)	9.0–11.5	0.10–0.60	0.4–0.8	≤0.20	—	0.50

Table 6. **Alloys of the Al–Zn system and their chemical compositions**

#	Alloy	Content, wt.%					
		Zn	Si	Cu	Mg	Others	Impurities (Fe, Cr, Ni, Ti, Mn, Zn, Pb, Sn), no more than
1	$AlZn_{10}Si_8Mg$	9.0–10.5	7.5–9.5	—	0.2–0.5	—	1.30
2	7050 [3]	6.0–6.4	—	2.2–2.3	2.2–2.3	Zr 0.10	Fe 0.15 Si 0.12
3	7055 [3]	7.6–8.4	—	2.0–2.6	1.8–2.3	Cr 0.04 Zr 0.08–0.25	Fe 0.15 Si 0.10 Mn 0.05
4	7075 [3]	5.1–6.1	—	1.2–2.0	2.1–2.9	Cr 0.18–0.28	Fe 0.50 Si 0.40 Mn 0.30
5	7150 [3]	5.9–6.9	—	1.9–2.5	2.0–2.7	Cr 0.04 Zr 0.08–0.15	Fe 0.50 Si 0.40 Mn 0.10
6	7475 [3]	5.2–6.2	—	1.2–1.9	1.9–2.6	Cr 0.18–0.28	Fe 0.12 Si 0.10 Mn 0.06

Table 7. **Alloys of the Al–Li system and their chemical compositions [3]**

#	Alloy	Content, wt.%								
		Li	Cu	Zn	Mg	Mn	Ti	Zr	Others	Impurities, no more than
1	2050	0.7–1.3	3.2–3.9	0.25	0.2–0.6	0.2–0.5	0.10	0.06–0.14	Cr 0.05	Fe 0.15 Si 0.12
2	2090	1.9–2.6	2.4–3.0	0.10	0.25	0.05	0.15	0.08–0.15	Cr 0.05	Fe 0.15 Si 0.10
3	2098	0.8–1.3	3.2–3.8	0.35	0.2–0.8	0.35	0.10	0.04–0.18	—	Fe 0.50 Si 0.40
4	2099	1.6–2.0	2.4–3.0	0.4–1.0	0.1–0.5	0.1–0.5	0.10	0.05–0.12	Cr 0.1–0.5	Fe 0.15 Si 0.12
5	2199	1.4–1.8	2.0–2.9	0.2–0.9	0.1–0.4	0.1–0.5	0.10	0.05–0.12	—	Fe 0.12 Si 0.10
6	8090	2.2–2.7	1.0–1.6	0.25	0.6–1.3	0.10	0.10	0.04–0.16	Cr 0.10	Fe 0.12 Si 0.10
7	2195 [9]	1.2	4.1	—	0.45	—	—	0.10	Ag 0.40	—

2. Techniques for Manufacturing Products from Aluminium Alloys

There is a wide variety of techniques for manufacturing parts, blanks, or semi-finished products from aluminium alloys. First of all, this is casting. Nevertheless, powder metallurgy, rolling (either alone or in combination with continuous casting) [10], hot and cold plastic deformation [11], and laser or friction welding [12] have been successfully implemented and are widely used. In recent years, additive manufacturing technologies have seen significant advancements: due to their low melting point, Al-based alloys are ideal, for example, for selective laser melting (SLM), *i.e.*, 3*D* printing of parts from powder mixes [13].

However, casting remains the primary manufacturing technology for aluminium alloy components. Regardless of the type of final processing (rolling [14, 15], hot forming [16], cold forming [11], *etc*.), the preliminary blank should be produced by casting.

The main advantage of casting is the wide variety of possible manufacturing techniques. For aluminium alloys, these include casting in disposable sand moulds, shell moulds, and metal moulds, as well as centrifugal casting, continuous casting, and a number of other techniques described in this section.

Products from all groups of cast aluminium alloys are produced using the technique of casting into a disposable sand mould. The main limitations of this technique are low production volume, reduced dimensional accuracy of castings, and surface quality, which necessitates larger machining allowances and a greater amount of machining [17].

Lost-wax casting ensures high dimensional accuracy and surface quality, as well as high productivity. However, this technique is more suitable for manufacturing parts made of special steels and alloys, as the production cost is quite high [17–19].

Production volumes of aluminium alloy castings can also be increased through lost-foam casting. The spread of this technique has been largely promoted by the research conducted at the Physico-Technological Institute of Metals and Alloys of the National Academy of Sciences of Ukraine under the supervision of O.Y. Shynskyi [20].

Casting into shell moulds made with hot-work tooling also allows for the production of parts from aluminium alloys. Compared to casting into metal moulds, this technique does not offer any particular technical advantages [17].

Gravity casting into metal moulds (coquilles) allows forming a fine-grained structure, which provides mechanical properties (strength and ductility) enhanced by 20–30%. In this case, machining allowances are reduced compared to disposable moulds, dimensional precision and surface quality of the castings are improved, and the leak-tightness is ensured [17].

Casting into graphite moulds is well suited for alloys with a wide solidification temperature range, such as Al–Cu alloys. However, the precision of the castings is lower compared to casting into metal moulds. Besides, there are significant limitations on the complexity of the shape, which makes it impossible to use this technique for the mass production of precision parts [17].

The results of Ref. [21] are significant from the viewpoint of forming the structure and properties of aluminium castings, depending on the mould material, as they show a drastic difference in the solidification dynamics between sand and metal moulds. Analytical calculations demonstrated the advantages of casting into metal moulds, particularly for alloys with a wide solidification range.

Die casting of aluminium alloys ranks second to gravity casting in moulds (coquilles) and is primarily used in the automotive industry [22]. In such castings, porosity is one of the most common flaws. For example, a study was conducted in Ref. [23] to determine the optimal conditions for an injection moulding production line using the $AlSi_9Cu_3$(Fe) alloy (EN AC-46500).

According to a 2024 study, the use of advanced aluminium alloys in die-casting is expected to grow by 8% annually until 2030, with demand led by the automotive and aerospace industries [22].

Compared to high-pressure die-casting, low-pressure die-casting and vacuum casting result in significantly fewer gas-related defects. These production techniques are also widely used on an industrial scale for aluminium alloys [24].

In Ref. [25], a rather innovative example of using high-pressure die casting to produce parts with virtually no gas- or air-related defects was presented. As noted, these manufacturing defects could be minimised only when the mould was filled under laminar flow conditions. Based on their overall properties, the parts can compete with those manufactured using other casting techniques.

The structure formation of cast components for aerospace applications made from $AlSi_9Cu_2$ alloy using high-pressure die-casting with laminar mould filling was investigated in another study. It was concluded that this technology is promising not only for producing cast blanks, but also for components suitable for further heat treatment [26].

Centrifugal casting of parts made from Al-based alloys has some limitations, such as the product shape (only bodies of revolution), contamination by non-metallic inclusions uniformly distributed in the castings due to the same densities of these inclusions and the metal, the segregation of heavier elements (*e.g.*, copper) due to centrifugal forces, and the formation of a porous ('banded') structure due to a wide solidification range of some alloys [17].

The die-casting process is crucial for the aerospace industry, but the range of products is quite specific: it includes fuselage panels, airframes of missiles and aircraft, *etc*. [17].

The most expensive and technologically complex technique is thixotropic casting (Thixo-Casting). In this case, the metal is injected into the mould in a semi-solid state, which ensures the absence of shrinkage and gas defects and the best mechanical properties [24, 27]. This technique is still in the research and development phase and is not yet ready for mass production of parts.

Continuous casting is widely used in the production of sheets and other semi-finished products made of aluminium alloys [10, 16]. The oldest and most conventional technique is double-strip casting, which is used for producing slabs with a thickness of 10 to 25 mm. The metal solidifies between two water-cooled rolls rotating in opposite directions. This technique is successfully used for alloys with a wide solidification range. The cooling rate during solidification is approximately 100 K/s. The castings did not contain shrinkage or gas-related flaws [10].

Currently, research into the possibilities of using recycled materials, including aluminium scrap obtained from military equipment destroyed during Russian armed aggression against Ukraine, is important. This type of recycling not only poses a significant environmental challenge but also offers opportunities for the restoration and development of resource-efficient technologies [28]. Studies have shown that the aluminium alloy produced by recycling military equipment was chemically similar to widely used conventional grades, such as 1925 and B95. Mechanical tests confirmed satisfactory performance of the alloy: UTS of up to 140 MPa and fracture elongation of up to 2.4%. Potential applications for this alloy include the manufacture of airframe components, cast engine parts, and other products that require a combination of high strength and corrosion resistance.

It was also shown in Ref. [29] that recycled materials obtained from military aluminium scrap cannot only be used but are quite suitable for producing new components. An approach was proposed for adjusting the chemical composition of a remelted alloy by adding zinc from scrap to get an Al–Zn–Mg composition. The as-cast alloy had the following mechanical properties: UTS of at least 140 MPa, and elongation of 5.4%, which corresponded to an intermediate stage before heat treatment.

However, as noted in Ref. [30], the use of recycled materials for the manufacture of critical components is still not possible without refining. For this purpose, a flux based on sodium chloride (NaCl) and potassium chloride (KCl) was proposed. The efficiency of the flux was evaluated by studying various types of aluminium scrap. It was conclusively proven that the flux efficiently removed non-metallic inclusions such as Al_2O_3, MgO, and SiO_2 from the melt, and partially hydrogen, thereby improving the quality of the alloy. The optimal flux concentration range (3–5%) provided maximum strength of up to 235 MPa and a fluidity of up to 585 mm due to effective refinement of the melt.

Over the past few years, the aviation industry has seen a rapid increase in the use of composite materials due to their low weight and enhanced mechanical properties. Composites are a relatively new type of structural material used in unmanned aerial vehicles, commonly known as drones. Additive manufacturing techniques make it possible to produce materials rapidly without the need for casting into moulds or tooling [31]. These techniques have high cost-effectiveness in prototype production, while their disadvantage is low production volume compared to casting.

Despite the widespread use of composite materials, aluminium alloys

Table 8. **Comparison of various techniques for manufacturing products from aluminium alloys**

Characteristics	Manufacturing technique															
	Casting technology															SLM
	Sand casting	Shell moulding	Lost-wax casting	Lost-foam casting	Graphite mould casting	Coquille casting	Casting into lined mould	High pressure die casting	Under pressure crystallisation casting or liquid forging	Squeeze casting	Under gas pressure or in vacuum	Thixocasting	Centrifugal casting	Continuous casting	Electro-slag remelting	
Reusability of casting mould	1	1	1	1	3	5	4	4	4	4	5	4	5	4	4	—
Complexity of part configuration	5	4	5	5	3	4	4	3	2	2	3	3	2	2	3	5
Minimum machining allowances	2	3	5	4	2	4	4	5	5	4	5	5	2	5	2	5
Surface cleanliness	1	3	5	4	3	4	4	5	5	4	5	5	3	3	4	4
Mechanical properties	3	3	2	4	3	4	3	4	5	4	4	5	5	4	5	4
Leak tightness (water tightness)	2	3	4	4	3	4	4	4	5	5	4	5	5	4	4	3
Performance	2	4	4	4	4	4	3	5	3	4	4	3	5	5	2	2
Manufacturing of small-sized parts	3	3	5	4	2	4	4	5	3	1	4	4	2	1	2	5
Manufacturing of large-scale components (dimensions in meters)	5	2	2	5	2	2	2	1	2	5	2	1	4	5	3	1
Manufacturing of parts with minimal wall thickness	3	3	5	4	3	4	4	5	3	5	4	5	3	3	3	5
Manufacturing of parts with holes	2	3	5	4	2	3	3	5	3	1	4	5	1	1	2	5
Cost	5	3	1	2	3	3	2	1	1	2	2	1	4	4	1	1

continue to be extensively researched and are playing an increasingly important role in the additive manufacturing of aircraft structural components [32].

Selective laser melting (SLM) enables the production of complicated 3*D*-structures with an unprecedented degree of freedom. Aluminium alloys are among the most promising materials for SLM production [33]. However, there are several challenges associated with laser melting, as Al-based alloys are prone to the formation of specific defects [13], which currently limit their potential applications.

The formation of defects such as cold shuts, voids, and cracks is a major issue in the selective laser melting (SLM) of Al alloys. The optimisation of SLM by adjusting the laser energy density can effectively eliminate cold shuts and porosity, thereby increasing the relative density. Cracks appear due to the rapid cooling during SLM, which results in stress concentration and the formation of columnar (coarse dendritic) grains. Microalloying with Sc, Zr, Ti, or the application of ultrasound promote the transition from columnar to rounded grains, eliminating the abovementioned flaws and enhancing the mechanical properties of Al-based alloys [34].

A comparative analysis of casting technologies and additive manufacturing of Al-alloy parts, including a relative assessment of each technique based on a number of parameters, is presented in Table 8. A score of 5 is excellent, while a score of 1 is unsatisfactory.

3. Methods of Alloying, Microalloying, and Modifying Aluminium Alloys

The fundamental difference between the concepts of ‘alloying’ and ‘modification’ is that, once added to the melt, the alloying element remains in the alloy both after solidification and after remelting.

The modifying element is completely removed after remelting. Due to various physical and chemical processes, the concentration of the modifying element in the melt gradually decreases after it is added, and after some time it disappears completely.

The ‘microalloying’ differs from ‘alloying’ in that the latter involves adding a relatively large amount of an element (1% or more), which is indicated in the alloy designation. Microalloying elements are added in quantities of hundredths or tenths of a percent and are usually not indicated in the alloy designation.

3.1. Alloying Elements That Determine Classification of Aluminium Alloys into Different Technological Groups

Silicon (Si) is a non-metal element with an atomic mass of 28 g/mol, a density of 2350 kg/m^3, and a melting point of 1430 °C.

The solubility of silicon in aluminium under normal conditions is of 0.05 wt.%, whereas it is of 1.65% at 577 °C. At this temperature, a eutectic transformation occurs. The silicon content in the eutectic mixture is 11.7% [35].

Magnesium (Mg) is an alkaline earth metal of the main group, with an atomic mass of 28 g/mol, a density of 1700 kg/m^3, and a melting point of 650 °C.

The solubility of magnesium in aluminium under normal conditions is 1.8 wt.%, whereas it is 17.4% at 449 °C. At this temperature, a eutectic transformation occurs. The magnesium content in the eutectic mixture is 33.0% [35].

Copper (Cu) is a main-group metal with an atomic mass of 64 g/mol, a density of 8920 kg/mi, and a melting point of 1083 °C.

The solubility of copper in aluminium under normal conditions is 0.1 wt.%, whereas it is 5.65% at 548 °C. At this temperature, a eutectic transformation occurs. The copper content in the eutectic mixture is 33.0% [35].

Zinc (Zn) is a main-group metal with an atomic mass of 65 g/mol, a density of 7100 kg/m^3, and a melting point of 419 °C.

Aluminium and zinc are mutually soluble. The solubility of zinc in aluminium under normal conditions is 2.0%, whereas it is 84% at 382 °C. At this temperature, a eutectic transformation occurs [35].

Magnesium is one of the most common alloying elements. Al-based alloys with magnesium are strengthened through the formation of the Mg_2Si phase. Besides, magnesium can participate in the formation of such phases as Al_2CuMg and $Al_5Mg_8Si_6Cu_2$. The partial dissolution of the Al_2CuMg phase during heat treatment, followed by the formation of fine precipitates enriched with Cu and Mg, is a key strengthening factor in aerospace aluminium alloys. The fine particles formed upon ageing effectively inhibit dislocation motion, thereby increasing the tensile and yield strengths. At the same time, sufficient ductility is preserved by means of eliminating brittle phase aggregates at the grain boundaries. The uniform distribution of reinforcing particles also increases fatigue life and corrosion resistance, which is critical for the long-term service of aircraft structures under altering loads and in aggressive environments.

The effect of microalloying with magnesium (0.09–0.23%) on an alloy containing 4.7% Cu, 0.33% Mn, 0.20% Ti, 0.14% Zr, and 0.17% V was investigated. Upon ageing, fine, dense, and evenly distributed Mg-containing precipitates form. The optimal Mg content is reported to be 0.09–0.13%, which increases the yield strength from 190 to 400 MPa and the tensile strength and UTS from 350 to 460 MPa, while fraction elongation drops from 16 to 2%. A positive effect on strength at elevated temperatures was also observed: the yield strength and UTS at 300 °C were 140 and 190 MPa, respectively, which exceeds the values for most similar alloys. Strength remained at the same level even after ageing at 300 °C for 1000 hours, provided that the magnesium content did not exceed 0.19%; otherwise, the *S*-phase (Al_2CuMg) formed, which reduced strength at elevated temperatures [36].

Zinc is added as an alloying element to increase the strength of aluminium alloys [37]. This element dissolves in the Al matrix and forms a supersaturated solid solution; after quenching and ageing, fine precipitates of the $MgZn_2$ η-phase form, which results in significant strengthening. Zinc is the main alloying element in high-strength alloys of the 7000 type (Al–Zn–Mg).

The addition of zinc into Al–Cu alloys can also significantly affect microstructure and mechanical properties. The presence of Zn changes the precipitation kinetics in the Al–Cu system, in particular by influencing diffusion and the stability of metastable θ′ precipitates, as well as by refining these precipitates and promoting their uniform distribution within the matrix. In contrast to Al–Zn–Mg alloys, where the main strengthening effect is associated with the η- and η′-phases, zinc in Al–Cu alloys primarily acts as a modifying and auxiliary element, enhancing the conventional strengthening due to the Al_2Cu phase. This combined effect makes it possible to better control the strength/ductility ratio, which is important for aircraft structures operating under alternating mechanical and thermal impacts [38–40].

The development of Al–Mg–Zn alloys was discussed in Ref. [39], where the modification of the chemical composition through a high Zn content and additional Cu alloying was considered. The work focused on the investigation of dispersion strengthening and corrosion resistance. It was shown that a combination of high Zn and Cu contents was critical for optimising the density and size of intermetallic precipitates and, consequently, for enhancing mechanical properties.

3.2. Alloying Elements That Do Not Affect Classification of Alloys but Are Indicated in Their Designations

Nickel (Ni) is a transition metal with an atomic mass of 59 g/mol, a density of 7900 kg/m^3, and a melting point of 1455 °C. It does not dissolve in aluminium, but forms a eutectic with a melting point of 640 °C and a nickel content of 5.7% [35].

Many researchers consider nickel effective in enhancing the strength of aluminium alloys at elevated temperatures. For example, the authors of [41] noted that the following five intermetallic compounds stable at high temperatures could form in Al-based alloys: Al_3Ni, Al_3Ni_2, AlNi, Al_3Ni_5, and $AlNi_3$.

In Al–Cu alloys, nickel is also considered an effective element that enhances thermal stability. In addition to the precipitation of the abovementioned phases, nickel promotes the formation of thermally stable Cu–Ni-containing compounds. This behaviour is quite different from the effect of the conventional strengthening phases in the Al–Cu system (θ, θ′), which coarsen significantly at elevated temperatures, leading to a decrease in strength. Thus, Ni promotes retaining the strengthening effect under prolonged thermal impact [42–44].

According to Ref. [45], the addition of nickel into Al–Si–Mg alloys significantly enhances the strength. However, Al–Ni binary alloys (with a nickel content of 1–5%) without other alloying elements have insufficient strength and have therefore not found widespread application. According to Ref. [45], the addition of nickel to Al–Si–Mg alloys significantly increases their strength. However, Al–Ni binary alloys (with a nickel content of 1–5%) without other alloying elements have insufficient mechanical strength and have therefore not found widespread application.

In Ref. [46], the effect of adding 0.5–4.0% Ni into an alloy containing 8.4–8.5% Cu, 2.2–2.3% Ce, 0.95–1.05% Mn, and 0.18–0.22% Zr was investigated. A positive effect on strength at 350 °C was confirmed for nickel content below 0.5%, whereas any addition of Ni was not effective at 400 °C. Under normal conditions, Ni did not noticeably affect the properties of the alloys, but it reduced strength when its content exceeded 2.0%.

Manganese (Mn) is a transition metal with an atomic mass of 55 g/mol, a density of 7400 kg/m^3, and a melting point of 1240 °C.

Manganese is insoluble in aluminium under normal conditions; its solubility at 657 °C is 1.8%. At this temperature, a eutectic transformation occurs. The manganese content in the eutectic mixture is about 4%; at higher concentrations, it forms a peritectic mixture with aluminium [35].

The addition of Mn and Cr to aluminium alloys plays a key role in providing microstructural stability. Manganese effectively neutralises the harmful effect of iron impurities by promoting the formation of a more morphologically ductile α-$Al_{15}(Fe, Mn)_3Si_2$ phase, which replaces the needle-shaped β-Al_5FeSi phase and thereby reduces the risk of brittle fracture. In Al–Cu–Mg alloys, manganese promotes the precipitation of the $Al_{20}Cu_2Mn_3$ phase at grain boundaries [47]; this impedes grain migration and recrystallization.

Adding 0.72% Mn to an experimental alloy containing 1.0% Si, 0.8% Mg, and 0.2% Fe increases the yield strength under both normal conditions and at elevated temperatures.

Titanium (Ti) is a transition metal with an atomic mass of 48 g/mol, a density of 4500 kg/m^3, and a melting point of 1670 °C. Titanium is insoluble in aluminium under normal conditions; its solubility at 665 °C is 0.28%. A peritectic mixture forms at this temperature [35].

The alloying of Al-based alloys with various Ti contents is rather common. In particular, adding Ti makes it possible to produce high-density Al–Cu alloys without cracks [48]. In an Al–Cu alloy without Ti, hot cracks always form along the coarse columnar grains due to high stress concentrations. The elimination of solidification cracking by means of Ti addition is attributed to a reduction in grain size and a change in grain morphology from columnar to equiaxed. According to [49], adding titanium to Al–Li–Cu alloys also significantly improves the grain structure, eliminates defects, and provides UTS of 516 MPa and an elongation of 6.8%.

In Refs. [50, 51], a similar effect of adding titanium as nanoparticles (0.7 wt.%) or an alloying element (0.3 wt.%) was studied. A transition from columnar to equiaxed grain structure was observed, which allowed for the complete elimination of hot cracks and anisotropy in the properties of Al 2024 and $AlSi_5Cu_2Mg$ alloys. This effect was achieved due to the precipitation of Al_3Ti nanoparticles, which were additional nucleation sites for solidification and improved the filling of interdendritic gaps with the melt [51].

Al_3Ti nanoparticles increase strength at elevated temperatures. During prolonged heating (100 hours at 300 °C,), stress fields form around these particles, which block dislocation motion and compensate for the matrix softening caused by the coarsening of the strengthening θ′-phases [52]. Therefore, the alloying with Ti ensures the stability of the fine-grained structure and the thermal stability of aluminium alloys over a wide temperature range.

The formation of strengthening phases also depends on the cooling rate and external impacts on the melt. A higher cooling rate (300 °C/h) accelerates reaching the peak hardness by activating additional nucleation sites for solidification [53]. At the same time, high-intensity ultrasonic treatment (UST) of the melt allows the fragmentation of Al_3Ti intermetallic precipitates, transforming them from coarse dendritic plates into fine compact particles [54].

The synergistic effect of alloying elements plays a key role in providing performance. Due to the combined microalloying with Ti and Zr, the Al_3(Zr, Ti) particles promote the heterogeneous nucleation of Cu-containing phases at the early stages of ageing [55]; besides, Mn addition leads to forming complex θ′-particles that remain stable above 300 °C and significantly enhance the strength at elevated temperatures [56].

In Al–Ti–B systems, the solidification is accelerated by the formation of a stable monolayer of the Al_3Ti phase on the surface of TiB_2 particles, which is an active nucleation site when free titanium is present in the melt [57].

Cadmium (Cd) is a transition metal with an atomic mass of 112 g/mol, a density of 8650 kg/m^3, and a melting point of 321 °C. This element is insoluble in aluminium under normal conditions, whereas it forms a monotectic with 6.7% cadmium at 649 °C [35].

In Ref. [58], the Al–Cu–Cd system was studied. It was shown that cadmium is an effective alloying element that significantly increases the strength of aluminium alloys (Al–Cu or Al–Si–Mg–Cu). Adding 0.2–0.3% Cd into the alloy led to the precipitation of finer strengthening particles upon ageing. This increased the strength of Al–Cu alloys after T6 heat treatment by 100–150 MPa.

However, cadmium is toxic, especially during the melting [59]. The authors of [60] and some researchers in previous years have shown that similar quantities of elements adjacent to Cd in the periodic table (Ag) provide a similar strengthening effect. However, this significantly increases the cost of the alloy, and, therefore, a suitable replacement for cadmium has not yet been found.

3.3. Alloying and Modifying Elements Not Included in Alloys' Designations

Lithium (Li) is a main-group metal with an atomic mass of 7 g/mol, a density of 530 kg/m^3, and a melting point of 180 °C.

The solubility of lithium in aluminium under normal conditions is of 0.70 wt.%; it is of 5.2% at 602 °C. At this temperature, the eutectic transformation occurs. The lithium content in the eutectic mixture is of 7.8% [35].

In the aviation industry, some researchers [3, 61, 62] focus on alloying with lithium to minimise the density of aluminium alloys. In the Li-containing Al–Cu alloys, the phases of Al_2CuLi type precipitate. As a result, lithium significantly increases the strength, but there are issues during the melting due to its excessively high chemical reactivity. The reduction in the density of these alloys is negligible due to low lithium content (no more than 3%). Adding 5% of lithium reduces the density by only 3%: as this element has a high cost, these alloys are the most expensive of all Al alloys.

Zirconium (Zr) is a transition metal with an atomic mass of 91 g/mol, a density of 6500 kg/m^3, and a melting point of 1850 °C. The solubility of zirconium in aluminium under normal conditions is 0.22 wt.%, whereas it is 0.28% at 660 °C. At this temperature, a peritectic transformation occurs [35].

Zirconium is considered one of the most effective microalloying elements, as it provides dispersion strengthening and high thermal stability up to 400 °C. In the Al_3Zr intermetallic compound, the diffusion is slow; however, this compound is metastable [35].

Adding 0.2% Zr to the alloys with 5–9% Cu and 0.4–0.5% Mn [63] improves the strength at elevated temperatures. At 300 °C, the strength is 150 MPa, which is comparable to other alloys.

As shown in Ref. [64], zirconium can be effectively added into aluminium alloys in the form of zirconium dioxide or zirconium tetrafluoride. The drawbacks of enriching aluminium with zirconium contained in master alloys were also noted. The use of ZrF_4 or ZrO_2 eliminated the need to produce pure zirconium and Zr-containing master alloys. It was found that the developed technique resulted in recovering zirconium from fluxes, and this element completely dissolved in the aluminium alloy.

In Ref. [65], the features of the strengthening of $AlSi_7$ alloy with zirconium added as part of tetrafluoride were studied. The efficiency of a complex flux containing zirconium tetrafluoride was proved. A mechanism for the zirconium refinement was described. As shown, the complex flux should be added to the molten aluminium alloy before other alloying elements such as silicon and magnesium are introduced into it. Modifying the $AlSi_7$ alloy with zirconium resulted in the refinement of the α-phase grains; no other structural changes were observed.

The effect of a zirconium tetrafluoride-based flux on the properties of the $AlCu_5Zr$ alloy was investigated in Ref. [66]. The melting of the alloy with a complex flux consisting of a mixture of zirconium tetrafluoride and lithium fluoride (mass ratio of ZrF_4 to LiF of 1:1.2), followed by T5 heat treatment, was studied. It was shown that the addition of Zr into the melt led to the formation of the Al_3Zr intermetallic precipitates, which acted as solidification sites and inhibited grain growth, thereby refining the microstructure. T5 heat treatment promoted dispersion hardening of the alloy due to the formation of fine $CuAl_2$ inclusions and increased structural homogeneity.

Scandium (Sc) is a main-group metal with an atomic mass of 45 g/mol, a density of 3000 kg/m^3, and a melting point of 1541 °C [35]. The solubility of scandium in aluminium at 655 °C is 0.27%, while it is of 0.07% at 527 °C.

Adding 0.2% Sc to the alloy containing 4.0% Cu and 0.8% Mn not only increases strength under normal conditions but also provides high strength at 150–300 °C. This value exceeds 200 MPa, which is higher than that of alloys with a higher (5–8%) Cu content, alloyed with 0.15–0.20% Zr, 0.15 or 1.0% Fe, and 0.5–1.5% Ni [63].

Combined alloying with 0.1% Sc and 0.1% Zr results in a significant synergistic effect [38, 67]. The Al_3Zr intermetallic compound precipitates due to the lower mobility of Zr atoms and acts as a diffusion barrier that inhibits coalescence [67, 68].

In Ref. [4], positive results were also achieved by adding 0.20% Zr and 0.13% Sc into an alloy containing 4% Cu and 1.2% Mg. The alloy strength increased from 290 MPa to 310 MPa, while the ductility remained high (18%).

More impressive results were obtained in Ref. [69]. Adding 0.10% Zr and 0.10% Sc into an alloy containing 1.0% Si and 1.0% Mg increased strength from 330 to 390 MPa under normal conditions, and from 140 to 170 MPa at 250 °C.

In Ref. [70], the effect of the Zr to Sc ratio in an alloy with 7% Mg was investigated. At Sc and Zr contents of 0.09% and 0.27%, respectively, and at equal amounts of these two elements (approximately 0.20%), strength of 420–440 MPa was reached at relatively high ductility (16–18%). When the Sc content (0.32%) exceeded that of Zr (0.11%), the strength increased significantly (up to 560 MPa), while the ductility remained high (10%).

The combined addition of Sc (0.40%) and Zr (0.34%) into the 2219 alloy (chemical composition shown in Table 3) increased the strength at room and elevated temperatures: at 20 °C, from 455 to 536 MPa; at 200 °C, from 248 to 306 MPa; at 250 °C, from 200 to 231 MPa. As is often the case with alloying, ductility decreased (from 19–20 to 5%) [6].

The combined alloying with zirconium and other elements was studied in Ref. [71]. The authors proposed to add 0.10–0.20 wt.% of Zr and Ti and perform isothermal ageing at 375–425 °C. The strength increased due to the precipitation of Al_3Zr or $Al_3(Zr_{1-x}Ti_x)$ nanoparticles. After 1000 hours of heat treatment, the Vickers microhardness increased from 250 to 450 MPa, when 0.10% Zr and Ti were added, and up to 700 MPa at 0.20% Zr and Ti.

Three elements (0.15% Zr, 0.21% V, and 0.17% Ti) were added to an alloy containing 4.7% Cu and 0.35% Mn. Then, the alloy was heat treated as follows: two-stage annealing (495 °C for 2 h + 528 °C for 10 h), and ageing at 200 °C for 4 h. As a result, Al_3(Ti, Zr, V) compounds formed in a total amount of approximately 0.20%. The strength increased from 375 to 410 MPa, while the elongation decreased from 12.1 to 10.6%. At 300 °C, the tensile strength and UTS increased from 140 to 155 MPa, while the elongation decreased from 30 to 24%. The authors determined the contribution of two factors to the increase in strength: a reduction in grain size accounts for only 10%, while 90% was related to the abovementioned intermetallic precipitates, besides the traditional θ′- and θ″-phases [72].

According to Ref. [73], an Al–Cu alloy containing Mn and Zr withstands prolonged exposure to temperatures up to 350 °C. In this alloy, the segregation of Mn and Zr at the interfaces between metastable phases played a significant role in their stabilisation.

Adding two elements (0.48% Zr and 0.33% Nb) to an alloy containing 6.5% Cu increased its strength under normal conditions from 440 up to 530 MPa, and from 245 up to 306 MPa at 200 °C [5].

Adding two elements (0.80% Sc and 0.45% Mg) to an alloy containing 6% Cu increased its UTS at room temperature from 441 up to 575 MPa, with practically no effect on ductility (4–5%). At 200 °C, the UTS increased from 238 up to 442 MPa, while the ductility decreased from 15 to 7% [7].

Adding two elements (0.18% Zr and 0.18–0.29% Sc) to the 3004 alloy (1.2% Mn, 0.6% Fe, 1.0% Mg, 0.25% Si) enhanced its properties. Under normal conditions, UTS increased 1.8-fold; at 300 °C, it rose from 62 to 78 MPa; at 375 °C, it reached 80 MPa [74].

Argentum (Ag) is a main-group metal with an atomic mass of 108 g/mol, a density of 10500 kg/m^3, and a melting point of 960 °C. The solubility of argentum in aluminium under normal conditions is 0.70 wt.%, whereas it is 48% at 566 °C. At this temperature, a eutectic transformation occurs. The argentum content in the eutectic mixture is 73.5% [35].

The authors of Ref. [60] and some researchers in previous years have shown that Ag provided a strengthening effect compared to that of the relatively toxic Cd. However, this significantly increased the cost of the alloys.

Tin (Sn) is a transition metal with an atomic mass of 119 g/mol, a density of 7300 kg/m^3, and a melting point of 232 °C.

The solubility of tin in the solid aluminium is very low (0.04%). Tin does not form any chemical compounds with aluminium, but it forms a eutectic structure with a melting point of 228 °C and a tin content of 99.5% [35].

In some cases, Sn was added to the aluminium alloy with 4% Cu. A positive effect on hardness was observed after artificial ageing at 200 °C for 3 h: at 0.04–0.06% Sn, the *HV* hardness increased from 150 to 360 MPa [75].

Ferrum (Fe) is a transition metal with an atomic mass of 56 g/mol, a density of 7870 kg/m^3, and a melting point of 1539 °C.

Ferrum is insoluble in aluminium under normal conditions; its solubility at 655 °C is of 0.052%. At this temperature, the eutectic transformation occurs. The Fe content in the eutectic mixture is 1.8% [35].

The positive effect of Fe on the strength of aluminium alloys at elevated temperatures was shown in Ref. [63]. At 300 °C, an alloy containing 7.5% Cu, 1.0% Mn, and 0.15% Fe had UTS of at least 100 MPa, while an alloy containing 5.0% Cu, 0.5% Mn, 1.0% Fe, and 1.0% Ni had UTS of at least 150 MPa.

Cerium (Ce) is the first element in the rare-earth metals group; it has an atomic mass of 140 g/mol, a density of 6770 kg/m^3, and a melting point of 800 °C.

Cerium is insoluble in aluminium under normal conditions; its solubility at 638 °C is 0.05%. At this temperature, the eutectic transformation occurs. The cerium content in the eutectic mixture is 13.0% [35].

In Refs. [76, 77], the effect of alloying with La and Ce was studied. It was shown that the mechanical properties of aluminium alloys at high temperatures depend largely on the amount and morphology of the intermetallic phases, which contain La and Ce. The $Al_{11}(Ce, La)_3$ and α-$Al_{15}(Fe, Mn)_3Si_2$ phases contribute to enhanced mechanical properties at high temperatures (300–400 °C). Adding more than 1% Ce and 1% La led to the formation of coarse, plate-like $Al_{11}(Ce, La)_3$ grains, which increases mechanical properties at 300 °C, but enhances them at 400 °C. Besides, cerium refines grains, modifies the eutectic structure, and acts as a degassing agent due to its reaction with gaseous (hydrogen) and liquid impurities.

In Ref. [78], some interesting findings regarding the strengthening of Al–Si alloys with more exotic rare-earth elements, such as Pr, Nd, Sm, Eu, Gd, Tb, Dy, Ho, Er, Tm, Yb, and Lu. The review [79] presented data on the structure formation in alloys around solidification centres formed by the aforementioned rare-earth elements; it was shown that the mechanism of REM action was similar to that of Ti, Zr, Nb, Ta, and some other elements. All these elements form intermetallic compounds of the Al_3Me type, which have lattice parameters similar to those of aluminium and are therefore effective modifiers of aluminium alloys of all groups.

Currently, these elements are rarely used in cast Al-based alloys, as they are very expensive and in short supply.

Molybdenum (Mo) is a transition metal with an atomic mass of 96 g/mol, a density of 10200 kg/m^3, and a melting point of 2620 °C. The solubility of Mo in aluminium under normal conditions is 0.1%; at the temperature of the peritectic transformation (707 °C), the solubility is 0.25%; it is 0.85% at 827 °C, and 1.9% at 927 °C [35].

In Ref. [80], molybdenum was added as a microalloying element into an experimental alloy containing 1.10% Mg, 1.50% Mn, 0.25% Si, and 0.30% Fe. The molybdenum content was 0.22%. As a result, the microhardness in the as-cast state increased from 62 to 68 *HV*, and after heat treatment, from 78 to 88 *HV*. The yield strength at room temperature increased from 184 to 215 MPa, and from 87 to 93 MPa at 300 °C.

3.4. Micro- and Nanomodifiers

TiN. It is produced by means of the chemical vapour synthesis (CVS) technique through the nitriding of titanium in an N_2 atmosphere by SHS at 1200 °C, the reduction of titanium from its oxide (at the same temperature), or PVD [81]. The melting point of this compound is of 2955–3700 °C, $HV = 14.5 \pm 1$ GPa, $HRC = 84$ [82].

The ultrafine particles of titanium nitride at fractions ranging from 0.008 to 0.014% are used to modify the $AlSi_7$ alloy. This enhances the strength, ductility, and impact toughness of the as-cast alloys and, to a greater extent, after the T5 or T6 heat treatments [83].

Besides, TiN particles at a fraction of 0.3% purify the metal matrix of impurities and strengthen it. Further modifying with titanium promotes the formation of elongated lamellar Al_3Ti precipitates [84].

The addition of ultrafine TiN particles (79–100 µm) to the melt of hypereutectic silumins reduces the grain size of primary silicon to 10–30 µm. The UTS increases from 250 to 270 MPa, and the *HB* hardness increases from 105 to 145 [85].

SiC compound is added into aluminium alloys as fine particles with a size of 1–100 µm in amounts of up to 20%; this increases hardness and wear resistance. As SiC has a high melting point, it is added into aluminium-based alloys as a master alloy at a fraction of 10–50% [86].

Adding 0.5% SiC nanoparticles into an Al–Cu alloy UTS and an elongation of approximately 4%, while adding 1.0–3.0% of these nanoparticles resulted in a UTS of 240–280 MPa and an elongation of 3–4% in the same alloys [87].

Si_3N_4 ultrafine particles were added in amounts of 0.008–0.014% to modify the $AlSi_7$ alloy; this increased strength, ductility, and hardness in the as-cast state; after the T6 heat treatment, the strength increased by 35% [83].

SiO_2 fine particles (0.1–500 μm) were added to modify the $AlSi_{12}$ alloy, which increased the strength, hardness, and ductility [88].

TiC nanoparticles, when added into the Al–Cu alloys, led to the following positive effects [87]:

- at a fraction of 0.7%, this provided UTS of 230 MPa, and an elongation of 1.5%;
- at a fraction of 5–9%, this provided UTS of 310–340 MPa, and an elongation of 8–11%;
- at a fraction of 10–30%, this provided UTS of 340–420 MPa, and an elongation of 2–5%.

TiB_2 + TiC nanoparticles, at a fraction of 0.1%, were added into an alloy containing 5.0% Cu, 0.6% Mg, and 0.5% Si. This allowed increasing the UTS under normal conditions from 435 to 517 MPa, at 180 °C from 357 to 413 MPa, and at 220 °C from 318 to 366 MPa. This modifier had virtually no effect on the ductility: under normal conditions, it was in a range of 17–22%; at 180 °C, of 16–18%; and at 220 °C, of 8–11%. Adding large amounts of nanoparticles (*e.g.*, 5–10% or more) was inappropriate [87].

Graphite nanoparticles were also used for modifying the $AlSi_{12}$ alloy. At a graphite content of 2.0%, the UTS was 140 MPa, the hardness was 50, and the ductility was 2.0% [88].

4. Heat Treatment of Aluminium Alloys

4.1. Various Heat Treatments and Their Effect on Microstructure and Properties of Alloys

For Al-based alloys, the heat treatments listed in Table 9 are predominantly used. The treatments with the same names vary in the heating temperatures, which are determined by the parameters of phase transformations in the alloys. Besides those treatments, special (proprietary) regimes are sometimes found in the literature, which may be similar to the basic treatments in general but have significant differences. A separate section at the end of this chapter considers the thermomechanical treatment (cold rolling + heat treatment or stress-relaxation heat treatment), which is also quite common for the cast alloys.

The review paper [89] can be considered fundamental for understanding the efficient heat treatments of aluminium alloys, particularly in the aviation industry. Heat treatment forms secondary phases, which harden the material. The thermal stability of these phases is crucial for the performance of materials at high temperatures. The phase coarsening and the competitive effects of different phases should be taken into account. Coarsening is a thermodynamically driven process caused by the tendency to reduce the total interphase energy, which consists of chemical energy and the associated elastic deformation energy between the secondary phase and the aluminium matrix. The use of computational methods (such as CALPHAD for calculating phase diagrams and VASP for calculating interphase energies) makes it possible to select alloying elements that can impede phase coarsening. Coarse phase components formed by impurities (Fe, Si) or primary phases that did not dissolve during heat treatment reduce the resistance to fatigue crack nucleation and propagation.

In Ref. [90], the key role of heat treatment in determining the mechanical properties of high-strength aluminium alloys was confirmed. The following standard heat treatments are used: T4 (provides maximum ductility), T5 and T6 (provide maximum strength), and T7 (a balance between strength and ductility). A distinctive feature is the use of high-temperature annealing at 320–380 °C.

The abovementioned heat treatment procedures are applicable, for example, for the Silafont®-36 alloy ($AlSi_9MgMn$). Since this alloy contains magnesium, the T5 heat treatment makes it possible to significantly increase the yield strength (by ≅100 MPa compared to the as-cast state), while preserving ductility. The T6 heat treatment provides maximum strength.

Table 9. **Heat treatments for cast aluminium alloys**

#	Heat treatment	Scheme	Quenching temperature*, °C	Artificial ageing temperature*, °C	Alloy type
1	T1	Artificial ageing	—	180 ± 5	Al–Zn
2	T4	Quenching + natural ageing	510 ± 5	—	Al–Si–Cu
			545 ± 5	—	Al–Cu
3	T5	Quenching + artificial ageing	535 ± 5	150 ± 5	Al–Si–Mg
			525 ± 5	175 ± 5	Al–Si–Cu
			545 ± 5	175 ± 5	Al–Si
4	T6, T64	Quenching + artificial ageing (incomplete artificial ageing for T64)	535 ± 5	175 ± 5	Al–Si–Mg
			525 ± 5	200 ± 5	Al–Si–Cu
			545 ± 5	170 ± 5	Al–Cu
5	T7	Quenching + incomplete artificial ageing	535 ± 5	225 ± 5	Al–Si–Mg
			525 ± 5	230 ± 5	Al–Si–Cu

Note: * Since the new EN and DSTU standards do not specify temperature ranges for the heat treatment of parts, the quenching and artificial ageing temperatures are taken from the previous DSTU 2839-94 standard

For the Magsimal®-59 alloy ($AlMg_5Si_2Mn$), the T5 heat treatment also increases the yield strength by 30%, but reduces the elongation by 30%. On the other hand, annealing reduces the yield strength by 30%, but increases the elongation by 30% [90].

Paper [91] is a general review of heat treatments used for high-strength aluminium alloys. For example, the T6 heat treatment is effective for achieving high yield and UTS in 2XXX and 7XXX series alloys. For instance, the yield strength of 345 MPa of an alloy in the as-cast state increased up to 689 MPa after the T6 heat treatment.

In Ref. [16], the effects of static and cyclic heat treatments on the Al 4043 alloy, which does not harden with heat treatment, were compared. Prolonged annealing at 200 °C promoted the softening of the matrix and the spheroidization of eutectic Si inclusions. This increased ductility, while preserving the strength at a certain level. However, repeated heating–cooling cycles (200 °C ↔ 25 °C) actually caused multidirectional diffusion of Si, which led to the formation of an irregular eutectic structure with Si particles of varying sizes. As a result, defects at the sharp ends of Si particles formed due to the difference in the thermal expansion coefficients of Al and Si, which reduced both strength and ductility. Therefore, these alloys are not capable of operating at high or elevated temperatures.

In Ref. [90], both conventional (T5, T6) and non-conventional heat treatments of cast aluminium alloys were considered. In particular, the T6 heat treatment (annealing and ageing) was recommended for components made of A356 and A380 alloys produced by high-pressure solidification (namely, liquid metal stamping [17]). After the T6 treatment, the materials had significantly higher tensile and yield strengths, as well as better ductility (6–7% elongation), compared to the same alloys after the T5 treatment (3–4% elongation).

In Ref. [92], the kinetics of phase precipitation and hardening in the Al–4% Cu alloy under various heat treatment conditions was investigated. Stepwise homogenization (at 450, 500, and 550 °C) followed by water quenching was performed. This was followed by natural ageing at 20 °C for 504 hours and artificial ageing at 180 °C. During the ageing, the supersaturated solid solution decomposed, which led to the sequential formation of metastable phases, including the Guinier–Preston (GP) zones, the θ″- and θ′-phases. The activation energies of the GP zones and θ″-phase formation (of approximately 60 kJ/mol) indicate that vacancy migration is the key driving force at the early stages of ageing. Microhardness reached its maximum (of approximately 110 *HV*) after 30 hours of artificial ageing. The maximum mechanical properties were achieved through the synergistic contribution of the θ″- and θ′-phases, with the dominant contribution of the θ″-phase.

In Ref. [60], the importance of multistage heat treatment for cast high-strength Al–Cu alloys was shown. In particular, it was emphasised that high-temperature annealing (homogenization) is necessary to eliminate the intracrystalline segregations that form upon solidification. A treatment similar to T6 was proposed, which includes two-stage homogenization (annealing) followed by water quenching and subsequent artificial ageing.

In Ref. [93], dilatometric analysis was used to study the phase transformations during heat treatment. When the Al–10% Cu alloy was heated, peaks were observed that corresponded to the sequential formation of metastable phases: θ'' (around 200 °C), θ (around 320 °C), and the stable θ-phase (around 400 °C).

The following ageing temperatures were recommended: 190 °C for the Al–Cu alloys, and 220–240 °C for the alloys containing Zr, Ti, or both these elements [94].

In Ref. [43], the effect of heat treatment on the phase transformations in the Sc-containing alloys was considered. At elevated temperatures (typically of 200–400 °C), the heat treatment led to the formation of nanoscale Al_3Sc precipitates with uniform distribution, which provided strengthening and played an important role in inhibiting recrystallization.

In Ref. [95], the evolution of the $AlSi_{4.5}Cd$ alloy structure during the heat treatment was studied on the example of blanks manufactured by casting under pressure. Even after casting with this technique, the presence of strengthening phase particles in the alloy matrix was a key feature. The basis for dispersion strengthening lies in the microheterogeneity within the grains containing α solid solution, which effectively increases the strength of the alloy at elevated temperatures. In the studied alloy, both solid solution and dispersion hardening work simultaneously.

The so-called 'Orovan strengthening' $\Delta\sigma_{Or}$ (or the Orovan mechanism) is one of the main mechanisms of material strengthening, particularly in dispersion-strengthened alloys. In this case, dislocations bend around immobile particles of another phase (in aluminium alloys, these are intermetallic precipitates), leaving dislocation loops around them, which increases the strength.

As could be expected, the Orovan strengthening is inversely proportional to the size of particles. For example, the strength decreases when the $Al_3(Sc_xZr_{1-x})$ particles coarsen. The annealing temperature affects the size of $Al_3(Sc_xZr_{1-x})$ particles. Annealing above the optimum temperature led to the coarsening of the particles, so the yield and tensile strengths decreased. As the annealing temperature grows from 230 to 500 °C, the average particle diameter increases from 15.2 ± 5.6 to 25 ± 7.6 nm. As the size of $Al_3(Sc_xZr_{1-x})$ particles in the alloy under study increased, the Orovan strengthening effect dropped from 51.1 MPa at 230 °C to 35.2 MPa at 500 °C [42, 96].

In Ref. [97], the mechanisms for stabilising hardening precipitates during heat treatment at elevated temperatures were studied. Annealing at 300 °C caused the coarsening of θ′-Al_2Cu and Al_3Sc precipitates. Combined microalloying with Sc and Zr provided the stabilisation of both types of precipitates, which resulted in a minimal rate of the θ′-phase coarsening. The slowing of the coarsening kinetics was attributed to a decrease in interfacial energy caused by the segregation of the alloying elements (Sc, Zr, Si) at the θ′-Al_2Cu/α-Al interphase boundaries.

In Ref. [44], the thermal stability of the θ′-Al_2Cu phase in the zirconium-microalloyed Al–Cu–Mn–Zr (ACMZ) alloy was studied under prolonged exposure to high temperatures (5000 hours at 300 °C). The θ′-precipitates did not coarsen, and the hardness of the alloy remained stable. The complete inhibition of the coarsening was attributed to the kinetic effects of Mn and Zr segregation at the interphase boundaries, which dominated over the thermodynamic factors. Based on this, it was stated that the classical theory of coarsening significantly overestimates the expected increase in precipitate size and fails to account for the stability provided by microalloying and heat treatment.

In Ref. [42], the role of heat treatment in the formation of the strengthening θ′-phase in the Al–Cu alloys was emphasised. Traditionally, preliminary cold working (such as stretching up to 6%) is used to promote the precipitation of this phase, which increases strength but reduces ductility and fatigue life. An alternative mechanism involves the heat treatment (ageing) of alloys containing Sc- and Zr-containing particles (Sc-enriched cores and Zr-enriched shells); this allows these particles to be used as effective heterogeneous nucleation sites for the θ′-phase precipitation.

In Ref. [98], the multistage homogenization was considered, which promoted the formation of nanoscale Al_3(Sc, Zr) precipitates in an Al–Cu–Sc–Zr alloy. These particles inhibited recrystallization, which resulted in the preservation of the deformed microstructure. The Al_3(Sc, Zr) precipitates acted as heterogeneous nucleation sites for the strengthening q″- and θ′-phases. This not only increased the strength but also enhanced the resistance to pitting corrosion by reducing the size of the precipitates.

In Ref. [38], the effectiveness of homogenization as a preliminary stage for heat treatment was shown. After this treatment, the volume fraction of the microsize constituents (eutectic phases) that formed during casting was significantly reduced. For example, their volume fraction in the Al–Cu alloy decreased from $2.13 \pm 0.20\%$ to $0.51 \pm 0.07\%$; in the Al–Cu–Sc–Fe alloy, it decreased to $0.23 \pm 0.05\%$.

In Ref. [67], the effect of heat treatment on the structural stability of the Al–Cu–Mg alloys was investigated. Annealing at 340–400 °C provided high structural stability. The contribution of T-phase particles to the strength remained stable within the abovementioned temperature range, as the changes in the size of these inclusions were minimal.

In Ref. [99], the development of a new high-temperature Al–Mg–Zn alloy and its optimal heat treatment were considered. The heat treatment (annealing at 460–500 °C and artificial aging) was developed based on thermodynamic calculations of the Al–Mg–Zn phase diagram. The annealing resulted in the formation of a single-phase α solid solution, while ageing at the temperatures of the α/T two-phase region allowed for the formation of stable T-phase ($Al_6Mg_{11}Zn_{11}$) precipitates. At a low ageing temperature (200 °C), nanoscale particles of the T' metastable phase precipitated. Above 300 °C, a stable T-phase predominantly formed at the grain boundaries. A sample pre-aged at 200 °C for 1 hour had an extremely high yield strength (≅260 MPa at 200 °C) that exceeded that of comparable commercial alloys.

The authors of Ref. [94] proposed a natural ageing for the Al–Zn–Mg alloys. This ageing for more than 20 days resulted in the diffusion of Zn into the Al matrix, which stabilised the microstructure.

In Ref. [100], the effect of heat treatment on the mechanical properties of a cast Al–Mg–Zn–Cu alloy was investigated. The alloy was annealed at 470 °C for 24 hours to form a solid solution. Increasing the ageing temperature (from 120 to 175 °C) significantly reduced the ageing duration and increased the volume fraction of the homogeneously distributed T-phase precipitates and GP zones. The yield strength increased from 377 to 451 MPa, but the ductility dropped (from 13.8 to 5.2%). Ageing at 150 °C yielded optimal properties (yield strength of 441 MPa, elongation of 9.6%).

For comparison, a few examples of alloys produced using selective laser melting, which also require further heat treatment, are discussed below.

In Ref. [13], the effect of heat treatment on the structure and properties of the alloys produced by selective laser melting (SLM) was investigated. The microhardness of the 2024 alloy produced by SLM was higher compared to cast billets, but lower than that of the billets after the T6 heat treatment. This indicates that heat treatment is a key factor in controlling the properties of the materials produced even with advanced SLM technology. The examples show that a heat treatment can significantly alter microhardness, either increasing or decreasing it. For the $AlSi_7$ alloy, the microhardness decreased from 94 to 45 *HV* after artificial ageing at 300 °C. In the case of the $AlSi_7Mg$ alloy treated similarly, the hardness decreased from 124–133 to 76–78 *HV*, whereas ageing at 165 °C increases the hardness of this alloy to 115–150 *HV*.

It should also be noted that SLM technology involves extremely rapid cooling. For instance, the $AlSi_{12}$ alloy after SLM had yield strength of 260 MPa, UTS of 380 MPa, and low ductility (3%). After annealing, there was a significant decrease in the yield strength (95 MPa) and UTS (140 MPa), but a sharp increase in ductility (up to ≈15%). After the T6 heat treatment, intermediate values of strength and ductility were achieved (yield strength of 210 MPa, UTS of 284 MPa, ductility of ≈5%) [101].

In Ref. [14], the study focused on the effect of cold rolling, as a method of plastic deformation, on the microstructure and corrosion resistance of the 6061 alloy [14]. This alloy contains inclusions with a diameter of 1–3 nm, which were identified as the β″-phase coherent precipitates and partially coherent particles of the *Q*′-phase. The additional cold working (in particular, cold rolling with a 90% reduction in thickness) induced a high dislocation density, which provided the service properties, including corrosion resistance.

In Ref. [15], a thermomechanical treatment including significant preliminary deformation (80% cold rolling) followed by artificial ageing at 150° was studied. The authors named this treatment as P-T6. The strength increased by 35% compared to the T6 heat treatment (yield strength of 507 *versus* 380 MPa) with no loss of ductility. After the P-T6 treatment, very homogeneous and fine θ′-phase precipitates (instead of GP zones) formed, whose atomic stoichiometry differed from that of the θ′-phase. The deformation defects accelerated the kinetics of the hardening phase precipitation (compared to the T6 treatment) due to short-circuit diffusion.

In Ref. [7], the effect of heat treatment on the strength of cold-rolled 2219 alloy microalloyed with Sc and Mg was studied. After casting and cold rolling, this alloy had yield strength of 542 ± 22 MPa after natural ageing. Heat treatment at 200 °C for 200 hours reduced the yield strength to 495 MPa, while the strength was preserved. This was due to a synergistic effect: grain boundary strengthening by $Al_3Sc/Al_3(Sc, Zr)$ particles (Orovan strengthening) and θ′-phase precipitates.

In Ref. [102], the phase separation upon intense plastic deformation at room temperature (≈25 °C) followed by artificial ageing was investigated. The deformation caused the decomposition of the supersaturated solid solution, and Cu-enriched particles grew rapidly along grain boundaries. Ageing at 100 °C (without deformation) resulted only in the formation of small GP zones. This confirmed that deformation (not just the heat treatment) was a key factor in the alloy strengthening.

In Ref. [103], the Creep Age Forming (CAF), *i.e.*, ageing under mechanical loading, was analysed. A high load (180 MPa) caused initial plastic deformation (≈1.5%), which led to the homogeneous bimodal precipitation of GP zones and the θ″-phase particles, as well as relatively coarse particles of the stable θ′-phase (which nucleated heterogeneously on dislocations). The precipitation of the θ′-phase, which normally did not form at low temperatures (165 °C) without loading, significantly enhanced hardening. Unlike the θ″-phase, the θ′-phase significantly affected the strengthening of the alloy by the Orovan mechanism, which resulted in a greater hardening effect.

The study [104] focused on the microstructural changes in the alloys strengthened with secondary phases, particularly the Al–Cu–Li (AA2195) alloys after a preliminary deformation and further heat treatment. After a higher preliminary strain (*e.g.*, up to 15%), the T8 heat treatment resulted in an increased strength of 670 MPa at a relatively high ductility (up to 7.5%).

In Ref. [105], the effect of natural ageing on the strength and strain hardening in 7XXX series (Al–Zn–Mg) alloys was studied. Clusters of the atoms of alloying elements formed, which effectively inhibited the movement of dislocations. A prolonged natural ageing (1440 hours) provided the best mechanical properties of the alloy. An abnormally high strain hardening was observed, which was even higher than that of artificially aged alloys. Plastic deformation of these alloys resulted in a transition from the dissolution to the growth of clusters of dissolved atoms. The dissolution dominated during the initial stage of deformation (below 7%). At higher strains (13 and 20%), the clusters grew due to the generation of vacancies, which accelerated diffusion. These clusters effectively strengthened the alloy, whereas the atoms of alloying elements reduced the stacking fault energy and interacted with the mobile dislocations through deformation-enhanced diffusion.

4.2. Intermetallic Phases in As-Cast and Heat-Treated Aluminium Alloys

The Al_3Ni, $FeNiAl_3$, AlCuNi, and Al_6Cu_3Ni phases increase the strength of Al alloys at elevated temperatures and counteract the detrimental effect of Fe. This is most effective in the case of the Al_6Cu_3Ni phase, which forms a complex and branched framework that provides high creep resistance [35]. Besides, the Al_3Ni phase contributes to the dispersion strengthening of aluminium alloys [81].

The Al_3Fe and Al_5Fe_2 phases significantly reduce corrosion resistance. In the Al–Mg–Zn alloys, iron has the most detrimental effect on ductility. However, adding 0.25% Be results in the formation of a compact $Al_4Fe_2Be_5$ phase, which increases the ductility up to 7% [106].

The Al_3Sc phase effectively refines the grains; the lattice of this compound is similar in structure to that of Al [107].

The Mg_2Si phase increases the strength of the hypoeutectic Al–Si alloys, as it precipitates at the grain boundaries and disrupts the solid solution matrix. At the same time, this phase reduces the corrosion resistance [108].

The $MgZn_2$ phase in zinc-containing alloys, which precipitates upon ageing, is the main hardening factor. Besides this phase, the Al_2Mg_3Zn intermetallic compound can form [35].

The $CuAl_2$ phase (θ-phase) strengthens Al–Cu alloys and increases the strength at elevated temperatures by forming a hard framework along the dendrite boundaries [81, 109]. However, this brittle intermetallic compound usually reduces mechanical properties [110].

The Al_3Mg_2 and Al_4Mg_3 phases (β-phases) strengthen the alloys but negatively affect their ductility [109]. These phases have a body-centred cubic lattice and precipitate in the Al–Mg alloys [35]. They enhance the corrosion resistance of the alloys [108].

The Al_3Ti phase is a common intermetallic compound in aluminium alloys. This phase enhances the strength at room and elevated temperatures, as it forms a hard framework along the dendrite boundaries. This compound is considered a first-type modifier [109].

Since this compound has cathodic properties, it enhances the corrosion resistance of aluminium alloys. A less common compound Al_2Ti can also form, which affects the alloy properties in the same way [108].

The Al_6CuNi phase forms a hard framework that increases the strength of the alloys up to 300 °C [112].

The Al_3Cu_2Ni (δ phase) and Al_6Cu_3Ni (T-phase) also precipitate in aluminium alloys [111]. These phases solidify as branched intermetallic compounds, inhibit diffusion, and reduce deformation at elevated temperatures and under external load [111].

The $Al_2Mg_3Zn_3$ phase is also known as a T-phase; it increases the strength of the alloys and their corrosion resistance [108, 109].

The Al_2CrMg is an S-phase that precipitates in the Al–Cu alloys. It provides maximum strengthening effects at room and elevated temperatures [35].

The Al_9Fe_3Ni phase precipitates in the Al–Cu–Mg alloys that do not contain manganese. It counteracts the harmful effect of iron. Nickel is also added, which enhances the mechanical properties at room and elevated temperatures while reducing susceptibility to cracking [35].

The Al_6Mn, Al_4Mn, and Al_3Mn phases form in the alloys with a high content of manganese. They increase strength at elevated temperatures up to 300 °C [35].

The Al_2CuMg phase forms in the Al–Cu alloys containing more than 0.2% magnesium; it reduces strength at elevated temperatures.

5. Properties of Aluminium Alloys and Their Applications

The mechanical properties of alloys in various states (as-cast or after heat treatment), which were determined in accordance with Ukrainian standards and testing regulations, are listed in Table 10. The chemical compositions of the alloys and their designations correspond to the data in Tables 1–5.

According to the data listed in Table 10, the need for heat treatment of cast aluminium alloys has been significantly reduced compared to the previous standard (DSTU 2839-94). Nearly half of the alloys (21) are recommended for use only in the as-cast state. This is partly due to the significant increase in the production of parts cast under pressure, for which heat treatment is virtually impossible.

The mechanical properties of the 2000 series alloys, which are used primarily in the aerospace industry, are listed in Table 11 [3].

Table 10. **Mechanical properties of cast aluminium alloys**

#	Alloy	State (as-cast or after heat treatment)	Mechanical property*			
			Yield strength, MPa	UTS, MPa	Elongation, %	Brinell hardness, HB
1	$AlCu_4MgTi$ (EN 21000)	T4	200	300–320	5.0–8.0	90
2	$AlCu_4Ti$ (EN 21100)	T6	200–220	300–330	3.0–7.0	95
		T64	180	280–320	5.0–8.0	85–90
3	$AlSi_2MgTi$ (EN 41000)	as-cast	70	140–170	3.0–5.0	50
		T6	180	240–260	3.0–5.0	85
4	$AlSi_7Mg$ (EN 42000)	as-cast	80–90	140–170	2.0–2.5	50–55
		T6	180–220	220–260	1.0	75–90
5	$AlSi_7Mg_{0.3}$ (EN 42100)	T6	190–210	230–290	2.0–4.0	75–90
		T64	120–180	200–250	4.0–8.0	60–80
6	$AlSi_7Mg_{0.6}$ (EN 42200)	T6	210–240	250–320	1.0–3.0	85–100
		T64	210	290	6.0	90
7	$AlSi_7(Mg)$ (EN 42300)	T6	140–160	200–220	3.0–8.0	60–70
		T64	110	170	10.0	55
8	$AlSi_7MgMn$ (EN 42400)	as-cast	100	200	7.0	60
		T5	120	210	7.0	60
9	$AlSi_{10}Mg$ (EN 43000)	as-cast	80–90	150–180	2.0–2.5	50–55
		T6	180–220	220–260	1.0–2.0	75–90
10	$AlSi_{10}Mg(Cu)$ (EN 43200)	as-cast	80–90	160–180	1.0	50–55
		T6	180–200	220–240	1.0	75–80
11	$AlSi_9Mg$ (EN 43300)	as-cast	80–90	160–170	2.0–2.5	50–55
		T6	180–210	220–290	2.0–4.0	75–90
		T64	180	250	6.0	80
12	$AlSi_{10}Mg(Fe)$ (EN 43400)	as-cast	140	240	1.0	70
13	$AlSi_{10}MnMg$ (EN 43500)	as-cast	120	200	5.0	65
		T5	150	270	4.0	80
		T7	120	200	12.0	60
14	$AlSi_{11}$ (EN 44000)	as-cast	70–80	150–170	6.0–7.0	45
15	$AlSi_{12}$(a, b) (EN 44100, EN 44200)	as-cast	70–80	150–170	4.0–6.0	50–55
16	$AlSi_{12}$(Fe)(a, b) (EN44300, EN 44500)	as-cast	140	220–240	1.0	60–65
17	$AlSi_9$ (EN 44400)	as-cast	80–90	170–180	4.0–5.0	50–55
18	$AlSi_{10}Mn$ (EN 44600)	as-cast	100	240	10.0	70
19	$AlSi_6Cu_4$ (EN 45000)	as-cast	90–100	150–170	1.0	60–75
20	$AlSi_5Cu_3Mg$ (EN 45100)	T4	180	270	2.5	85
		T6	280	320	1.0	110
21	$AlSi_5Cu_1Mg$ (EN 45300)	T4	120–140	170–230	2.0–3.0	80–85
		T6	200–210	230–280	1.0	100–110
22	$AlSi_5Cu_3$ (EN 45400)	T4	110	230	6.0	75
23	$AlSi_7Cu_{0.5}Mg$ (EN 45500)	T6	190–240	250–320	1.0–4.0	85–100
24	$AlSi_7Cu_1Mg_{0.6}$ (EN 45600)	T7	260	290	2.0	100
		T64	250–270	280–320	3.0–4.0	100
25	$AlSi_9Cu_3(Fe)$ (EN 46000)	as-cast	140	240	1.0	80

The End of Table 10

#	Alloy	State (as-cast or after heat treatment)	Mechanical property*			
			Yield strength, MPa	UTS, MPa	Elongation, %	Brinell hardness, HB
26	$AlSi_{11}Cu_2$(Fe) (EN 46100)	as-cast	140	240	1.0	80
27	$AlSi_8Cu_3$ (EN 46200)	as-cast	90–100	150–170	1.0	60–75
		T5	130	190	1.0	80
		T6	250	280	1.0	95
28	$AlSi_7Cu_3Mg$ (EN 46300)	as-cast	100	180	1.0	80
29	$AlSi_9Cu_1Mg$ (EN 46400)	as-cast	90–100	135–170	1.0	60–75
		T6	235	275	1.5	105
30	$AlSi_9Cu_3$(Fe)(Zn) (EN 46500)	as-cast	140	240	1.0	80
31	$AlSi_7Cu_2$ (EN 46600)	as-cast	90–100	150–170	1.0	60–75
32	$AlSi_{12}$(Cu) (EN 47000)	as-cast	80–90	150–170	1.0–2.0	50–55
33	$AlSi_{12}Cu_1$(Fe) (EN 47100)	as-cast	140	240	1.0	70
34	$AlSi_{12}$(Fe) (EN 47200)	as-cast	140	240	1.0	65
35	$AlSi_{12}CuMgNi$ (EN 48000)	as-cast	140	240	1.0	90
		T5	185	200	1.0	90
		T6	240	280	1.0	100
36	$AlSi_{17}Cu_4Mg$ (EN 48100)	as-cast	180	200	1.0	90
		T5	260	295	1.0	125
37	$AlSi_{15}Cu_3MgFe$ (EN 48200)	as-cast	240	740	1.0	100
38	$AlMg_3$ (EN 51100)	as-cast	70	140–150	3.0–5.0	50
39	$AlMg_9$ (EN 51100)	as-cast	130	200	1.0	70
40	$AlMg_5$ (EN 51300)	as-cast	90–100	160–180	3.0–4.0	55–60
41	$AlMg_5Si$ (EN 51400)	as-cast	100–110	160–180	3.0	60–65
42	$AlMg_5Si_2Mn$ (EN 51500)	as-cast	140	250	5.0	70
43	$AlZn_{10}Si_8Mg$ (EN 71100)	T1	190–200	210–230	1.0	80–90

Note: *If a single value is indicated, the property should not be lower than that value

Table 11. **Mechanical properties of 2000 series alloys**

Alloy	Property		
	Yield strength, MPa	UTS, MPa	Elongation, %
2024	324	428	21
2026	365	496	11
2224	345	476	10
2324	370	475	8
2524	306	434	24

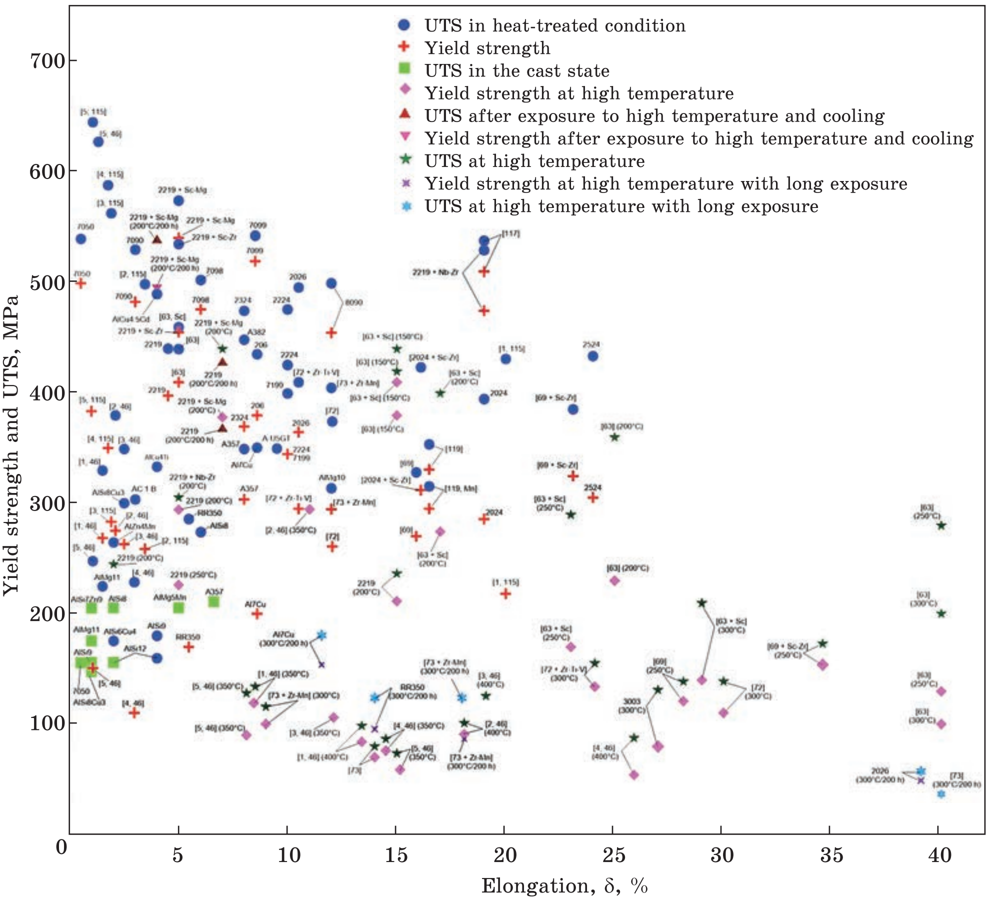

Fig. 1. Strength and ductility characteristics of standard and experimental aluminium alloys depending on experimental conditions

Table 12. **Mechanical properties of Al–Li alloys**

Alloy	Property		
	Yield strength, MPa	UTS, MPa	Elongation, %
2050	500	540	—
2090	483	531	3
2098	476	503	6
2099	520	543	8
2199	345	400	10
8090	455	500	12

The mechanical properties of the Al–Li alloys, which are also used in the aerospace industry, are listed in Table 12 [3].

Table 13. **Chemical compositions of experimental alloys**

Designation in Figs. 1, 2	Author's designation	Alloying element				
		Si	Cu	Mg	Mn	Zn
1	2	3	4	5	6	7
[113]	0Mn (base)	1.01	—	0.83	—	—
[113 + Mn]	0.72Mn	1.02	—	0.84	0.72	—
[80]	B	0.26	—	1.07	1.55	—
[80 + Mo]	M	0.26	—	1.10	1.54	—
[1, 114]	Alloy 1	—	5.08	5.18	—	4.02
[2, 114]	Alloy 2	—	10.48	8.33	—	7.17
[3, 114]	Alloy 3	—	4.06	0.99	—	1.26
[4, 114]	Alloy 4	—	9.73	1.83	—	2.70
[5, 114]	Alloy 5	—	17.34	1.80	—	2.48
[1, 115]	Alloy 1	—	4.34	—	—	0.45
[2, 115]	Alloy 2	—	12.04	—	—	1.88
[3, 115]	Alloy 3	—	10.56	—	—	0.59
[4, 115]	Alloy 4	—	19.04	—	—	1.72
[5, 115]	Alloy 5	—	18.68	—	—	0.42
[6, 115]	Alloy 6	—	18.01	—	—	0.47
[1, 116]	B	0.12	4.9	0.13	0.38	—
[2, 116]	Z2	0.10	4.9	0.11	0.36	—
[3, 116]	V2	0.10	4.8	0.13	0.38	—
[4, 116]	ZV	0.08	4.7	0.12	0.35	—
[5, 116]	SZ	0.09	4.4	0.11	0.33	—
[117]	Al–Zn–Mg–Sc–Zr	0.07	—	2.10	—	6.1
[118]	3003	0.6	0.05–0.20	—	1.0–1.5	—
[1, 74]	SZ0 (base)	0.24	—	0.97	1.23	—
[2, 74]	SZ15	0.25	—	1.04	1.18	—
[3, 74]	SZ30	0.25	—	1.01	1.19	—
[1, 46]	0Ni	—	8.46	–	0.98	—
[2, 46]	0.5Ni	—	8.53	—	1.03	—
[3; 46]	1.0Ni	—	8.45	–	1.02	—
[4, 46]	2.0Ni	—	8.49	—	1.05	—
[5, 46]	4.0Ni	—	8.34	—	0.96	—
[63]	Al–Cu–Mn	0.11	4.0	—	0.77	—
[63 + Sc]	Al–Cu–Mn–0.2Sc	0.10	4.1	—	0.74	—
[119]	1#	0.81	1.01	0.90	0	—
[119 + Mn]	3#	0.82	1.04	0.91	1.06	—
[69]	Al–Mg–Si	1.18	—	1.01	0.03	0.03
[69 + Sc–Zr]	Al–Mg–Si–Sc–Zr	1.13	—	1.05	0.04	0.02
[72]	base	0.06	4.7	0.14	0.35	—
[72 + Zr–Ti–V]	TEs	0.04	4.7	0.12	0.34	—
[73]	AC	—	5.3	—	—	—
[73 + Zr–Mn]	ACMZ	—	5.2	—	0.21	—

Notes: 1. The authors of Ref. [5] added 0.33% Nb to the well-known 2219 alloy (its chemi-
the diagram, the corresponding symbols are labelled '2219' and '2219 + Nb–Zr'. 2. In
0.15 to 0.34%. In the diagram, the corresponding symbols are labelled '2219' and '2219
position is given in Table 5) without additional alloying were studied after heat treat-
three conventional alloys (206, Al_7Cu, and RR350; their compositions are given in Tab
treatment were determined under normal conditions and at high temperatures. In the
[7], 0.80% Sc and 0.45% Mg were added to the well-known 2219 alloy (its chemical
'2219' and '2219 + Sc–Mg'. 6. In Ref. [4], 0.20% Zr and 0.13% Sc were added to the
corresponding symbols are labelled '2024' and '2024 + Zr–Sc'.

content, wt.%								
	Ti	Fe	Zr	Cr	V	Ni	Sc	others
	8	9	10	11	12	13	14	15
	0.018	0.22	—	—	—	—	—	—
	0.016	0.23	—	—	—	—	—	—
	—	0.28	—	—	—	—	—	—
	—	0.29	—	—	—	—	—	Mo 0.22
	—	—	—	—	—	—	—	—
	—	—	—	—	—	—	—	—
	—	—	—	0.97	—	—	—	—
	—	—	—	1.07	—	—	—	—
	—	—	—	1.10	—	—	—	—
	—	1.04	—	0.99	—	—	—	—
	—	0.99	—	0.83	—	—	—	—
	—	—	—	1.72	0.87	—	—	—
	—	1.01	—	0.79	—	—	—	—
	0.94	—	—	1.85	—	—	—	—
	—	—	—	1.85	0.82	—	—	—
	0.11	0.13	—	—	—	—	—	—
	0.10	0.13	0.18	—	—	—	—	—
	0.10	0.13	—	—	0.25	—	—	—
	0.09	0.13	0.15	—	0.13	—	—	—
	0.08	0.12	0.15	—	—	—	0.16	—
	0.015	—	0.14	—	—	0.025	0.12	—
	—	0.7	—	—	—	—	—	—
	—	0.60	—	—	—	—	—	—
	—	0.59	0.18	—	—	—	0.18	—
	—	0.57	0.17	—	—	—	0.29	—
	—	—	0.21	—	—	—	—	Ce 2.29
	—	—	0.17	—	—	0,48	—	Ce 2.33
	—	—	0.22	—	—	1.06	—	Ce 2.27
	—	—	0.19	—	—	1.91	—	Ce 2.22
	—	—	0.18	—	—	3.95	—	Ce 2.31
	—	0.05	—	—	—	—	—	—
	—	0.03	—	—	—	—	0.20	—
	—	—	0.20	—	—	—	—	Sr 0.023
	—	—	0.22	—	—	—	—	Sr 0.019
	0.02	0.05	—	0.02	—	—	—	—
	0.01	0.06	0.11	0.03	—	—	0.10	—
	0.05	0.12	—	—	—	—	—	—
	0.17	0.11	0.15	—	0.21	—	—	—
	—	—	—	—	—	—	—	—
	—	—	0.15	—	—	—	—	—

cal composition is given in Table 3), and increased the Zr content from 0.15 to 0.48%. In
Ref. [6], 0.40% Sc was added to the 2219 alloy, and the Zr content was increased from
+ Sc–Zr'. 3. In Ref. [2], the properties of the well-known A357 alloy (its chemical com-
ments. In the diagram, the corresponding symbols are labelled 'A357'. 4. In Ref. [8],
le 3) without additional alloying were studied. The properties of these alloys after heat
diagram, the corresponding symbols are labelled '206', 'Al_7Cu', and 'RR350'. 5. In Ref.
composition is given in Table 3). In the diagram, the corresponding symbols are labelled
well-known 2024 alloy (its chemical composition is given in Table 3). In the diagram, the

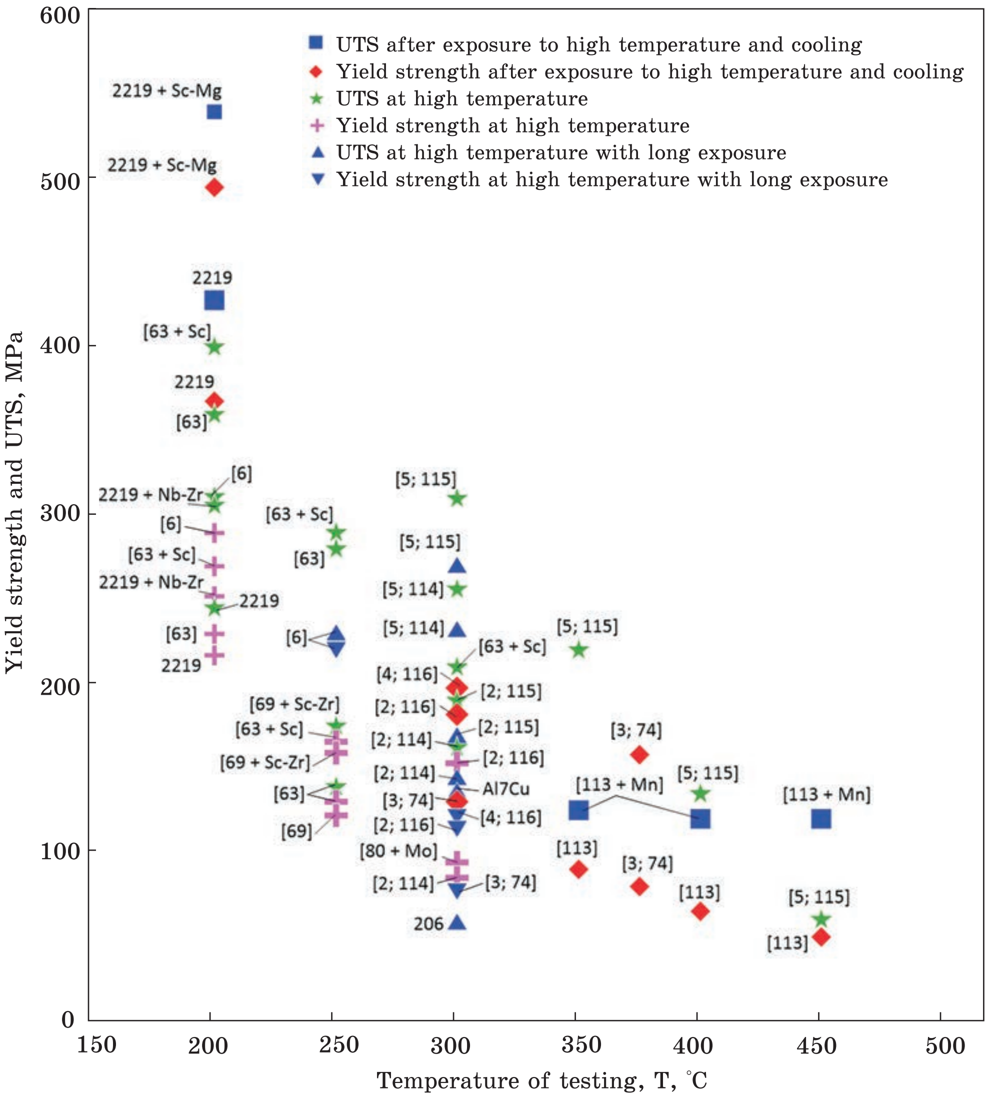

Fig. 2. Yield strength and UTS of aluminium alloys at high temperatures and different holding times

The chemical compositions of the experimental alloys, as reported in the scientific papers [46, 63, 69, 72–74, 80, 113–119], are listed in Table 13. Some alloys are not listed in this table because they are similar to the conventional alloys, but their chemical compositions were adjusted. Details regarding changes in the chemical compositions of those alloys, along with references to the original sources, are provided in the notes to Table 13.

The mechanical properties of all conventional and experimental alloys are compared in Fig. 1. On the diagram, each symbol represents a combination of the alloy strength and elongation. Both UTS and yield strengths are given; the vast majority of researchers pay considerable attention to the latter parameter. The mechanical properties at high temperatures are also given (the temperatures are indicated in parentheses above the experimental data points). When the properties were determined during prolonged exposure of the samples to high temperatures, the duration of exposure in hours is indicated in addition to the temperature.

The high-temperature properties of aluminium alloys, particularly, for aerospace applications, are the subject of extensive research. The scatter of these data is so vast that only a few examples of high-temperature properties are given on the abovementioned diagram. Therefore, Fig. 2 only shows the strength characteristics of the alloys, which depend on temperature and the experimental conditions; this allows for a comparison of the high-temperature strength of various alloys under different operating conditions.

An analysis of the information presented in Figs. 1 and 2 leads to the following conclusions.

(i) The mechanical properties of all alloys in the as-cast condition are always lower than those ones in the heat-treated states. This confirms the advisability and effectiveness of heat treatments under various conditions.

(ii) The addition of alloying or modifying elements usually enhances strength.

(iii) The addition of alloying elements, which increase strength, always results in a reduction in ductility to some extent.

(iv) The ratio between the yield strength and UTS varies greatly among different alloys. Therefore, there are no specific patterns or coefficients relating these two properties.

(v) The chemical elements that have the greatest effect on the mechanical properties are Zr, Ti, and Sc. These elements, along with nickel, increase the strength at elevated temperatures.

(vi) Aluminium alloys are capable of withstanding fairly high mechanical loads at elevated temperatures (150–350 °C, and in some cases up to 450 °C), even after prolonged exposure (200–300 hours). Therefore, these alloys are virtually indispensable for use in the aviation industry, where components work under extreme conditions.

(vi) At high temperatures, the strength of all alloys is significantly lower than at room temperature, while ductility increases considerably (in some cases by up to 40%).

(viii) In terms of UTS (400 MPa and above), a significant number of aluminium alloys are now fully capable of competing with such common cast materials as high-strength cast irons and unalloyed steels. Aluminium alloys have an advantage in strength at elevated temperatures.

6. Conclusions

Recent literature reviews confirm that aluminium alloys are among the most promising casting and structural materials. Moreover, in terms of the development rate of scientific and technological solutions aimed at improving their structure and properties, these materials currently hold a leading position.

The evolution of aluminium casting alloys reveals a clear transition from standard or commercial-grade to unique, multi-component compositions, which ensures a gradual increase in mechanical and special properties. Compared to early aluminium–silicon alloys, whose strength does not exceed 300 MPa with ductility within 5%, and aluminium–copper alloys, which achieve strength of up to 450 MPa and ductility of 10%, some modern alloys exhibit strengths exceeding 600 MPa and ductility levels of 20%. This makes them highly competitive with carbon steels and high-strength cast irons.

Considerable attention in the scientific literature is devoted to methods of alloying, modification, and heat treatment of aluminium alloys, with the number of publications on this topic growing annually. This indicates significant potential for improving the mechanical and special properties of these materials. The development of casting technologies, as well as physical and structural research methods, enables the study of previously inaccessible aspects of alloy structure formation and targeted control of these processes.

Current global scientific and technological trends are focused on further expanding the functional capabilities of aluminium alloys. Novel scientific and technological solutions contribute to the expansion of application areas for these unique materials. The most relevant directions for further research include studying the influence of alloying elements (such as Zr, Sc, Zn, Ni, Ti, and rare earth metals) on structure formation and properties, increasing the upper limit of operating temperatures of alloys to above 300 °C, and ensuring efficient processing of aluminium scrap and rational use of secondary raw materials.

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ЛИТІ АЛЮМІНІЄВІ СПЛАВИ ЯК КОНКУРЕНТОСПРОМОЖНІ СУЧАСНІ КОНСТРУКЦІЙНІ МАТЕРІАЛИ: ХІМІЧНИЙ СКЛАД, ВИГОТОВЛЕННЯ, ОБРОБЛЕННЯ, ВИПРОБУВАННЯ, МЕХАНІЧНІ ВЛАСТИВОСТІ ТА ЗАСТОСУВАННЯ

У статті наведено огляд сучасних алюмінієвих сплавів, технологій, що використовуються для виробництва й оброблення їх, а також формування їхньої мікроструктури та властивостей. Наразі машинобудування має тенденцію до широкого використання ресурсоефективних, екологічно чистих і високоточних технологій. Оскільки алюмінієві сплави мають низьку густину, відносно високу міцність, достатню пластичність і певні особливі властивості (достатньо високу міцність за підвищених температур), вони є ідеальними матеріалами для широкого використання в автомобільній, приладобудівній та аерокосмічній промисловостях. Наведено класифікацію алюмінієвих сплавів з різними групами на основі легувальних елементів. Проведено аналогії між національною (українською) класифікацією сплавів і тими, що використовуються в провідних промислово розвинених країнах. Проаналізовано подібності та відмінності в хімічному складі та властивостях сплавів. Розглянуто технології лиття й адитивного виробництва для виготовлення деталей з алюмінієвих сплавів. Показано вплив процесів структуроутворення під час затвердіння стопів на їхній фазовий склад і механічні властивості. Розглянуто різні легувальні елементи та проаналізовано вплив їх на мікроструктуру та властивості. Рекомендовано обмеження вмісту кожного легувального елемента й основні технологічні особливості топлення стопів. Також обговорено методи модифікування стопів. Показано вплив різних модифікувальних елементів на морфологію загальної структури та фази зміцнення. В окремому розділі статті наведено огляд різних термічних оброблень литих алюмінієвих сплавів. Ці сплави дуже чутливі до термічного впливу. Тому вибір відповідних умов термічного оброблення є ключовим для значного підвищення міцності (іноді в 1,5–2,0 рази) шляхом утворення додаткових структурних (фазових) складових. Зокрема, визначено процедури оброблення деталей, що використовуються в екстремальних умовах (температури понад 300 °C, агресивні середовища тощо, як в авіаційній або аерокосмічній промисловостях). Перераховано найпоширеніші інтерметалідні фази в алюмінієвих сплавах; обговорюється вплив кожної з цих фаз на властивості. Побудовано порівняльну діаграму, яка демонструє механічні властивості (міцність і пластичність) сплавів, що зазвичай використовуються, та тих, що розробляються (експериментальні сплави). Дані на цій діаграмі узагальнюють результати багаторічних досліджень науковців з різних країн, які були опубліковані у провідних наукових журналах. Також наведено дані щодо міцності за підвищених температур, розглянуто механічні властивості за високих температур (до 450 °C) і за тривалого впливу екстремальних умов. Визначено актуальні науково-технологічні проблеми, над якими нині працюють фахівці з лиття алюмінію.

Ключові слова: алюмінієві ливарні сплави, легування, термічне оброблення, механічні властивості, структуроутворення, утилізація.