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Mechanical behaviour of Zn–Al–Cu–Mg alloys: Deformation mechanisms of as-cast microstructures

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ABSTRACT

We study the effects of dilute Mg addition on the microstructure formation and mechanical properties of a ZnAl4Cu1 alloy. On the basis of the composition of the commercial alloy Z410 (4 wt% Al, 1 wt% Cu, and 0.04 wt% Mg), three laboratory alloys with different Mg contents (0.04 wt%, 0.21 wt% and 0.31 wt%) are characterised in terms of their mechanical properties and microstructures using ex-situ and in-situ tensile tests in conjunction with scanning electron microscopy (SEM) and electron backscatter diffraction (EBSD). Increasing Mg content causes the precipitation of Mg_2Zn_{11} phase precipitates and refined lamellar spacings in the eutectoid phase. The alloy with a medium Mg content (0.21 wt%) exhibits the highest yield strength both at room temperature and at elevated temperatures. Further, we show that dilute Mg alloying causes an improvement of the ductility of ZnAl4Cu1 base-alloys, especially at elevated temperatures. In addition, the alloys reveal two distinct deformation regimes distinguishable close to room temperature and at commonly employed strain rates, with work hardening and brittle fracture exhibited at room temperature and/or elevated strain rate ($5 \times 10^{-4}\ s^{-1}$), and work softening and ductile fracture at elevated temperature and/or low strain rate ($6 \times 10^{-6}\ s^{-1}$). The deformation mechanisms and fracture behaviour in both regimes are investigated and the underlying physical mechanisms of the observed phenomena are discussed.

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1. Introduction

Zinc in its unalloyed form is relatively soft and brittle at low and ambient temperatures. This is due to the low symmetry, high mechanical anisotropy of its hexagonal crystal structure and the limitation to mainly (0001) <11$\bar{2}$0> basal ⟨a⟩ dislocation slip and {10$\bar{1}$2}⟨10$\bar{1}\bar{1}$⟩ twinning at room temperature. These deformation systems alone do not offer enough independent slip and twinning systems to fulfil the von Mises criterion requiring at least 5 independent deformation systems for compatible polycrystalline deformation [1,2]. On the other hand, it was reported that due to the relatively low melting point of Zn, high temperature deformation systems, such as activation of prismatic and pyramidal slip systems, diffusion of point defects and grain boundary sliding, are readily activated at temperatures slightly above room temperature [1,3–5]. Zhang et al. [6] determined that the activation energy for plastic deformation in ultra-fine-grained Zn is close to the activation energy for grain boundary diffusion in pure Zn (60 kJ/mol) in the temperature range of 20–60 °C (59 kJ/mol). Moreover, the fracture mechanisms of pure Zn were investigated by several groups [1,4,7–10]. Hughes et al. [9] conducted miniaturised disc tests on polycrystalline Zn in the temperature range 77–328 K, and showed that the predominant cleavage plane was the (0001) basal plane. Moreover, he also observed a progressive increase of ductile fracture with increasing temperature. Liu et al. [4] investigated the fracture behaviour of polycrystalline zinc under tensile deformation at room temperature and observed cleavage fracture at high strain rates and at low strain rates failure from tearing cracks developed along grain boundaries, twin boundaries, kink boundaries and coalescing of cavities nucleated at triple junctions.

Zn–Al based alloys show attractive physical and mechanical properties [11–15]. Particularly due to their relatively low melting point (around 400 °C), corrosion resistance and excellent castability [16–19], Zn–Al based alloys are widely used for various engineering applications and predominately processed by die casting [20–24]. More specifically, they are used in production of both structural and decorative parts for automotive, electrical and electronic industries as well as general purpose machinery and equipment requiring high manufacturing precision [12,15,25–27]. However, Zn–Al alloys have only limited ductility at low temperatures [14,15,21,23,28] and suffer from a pronounced drop in

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strength at elevated temperatures [13,24,29]. To circumvent these drawbacks, many attempts have been made to strengthen Zn–Al alloys by the addition of alloying elements, such as Cu, Mg, Si, Ti, Ni, Mn, Zr and rare earth (RE) [14–16,18,19,23,25–27,30–37], heat treatment [12,30,38] and advanced metallurgical processes, such as continuous casting, thermal evaporation, rapid solidification and spray-deposition [17,39], the formation of ultrafine grained microstructures through severe plastic deformation [20,21,40–44] as well as particle reinforcement of non-metallic compounds [45–47]. The addition of alloying elements proved to be the most convenient and effective way to improve the mechanical properties of Zn–Al alloys. Alloying with Mg was reported to be effective in increasing the hardness and strength of Zn–Al alloys [25,27,32] and refining the lamellar eutectoid structure of Zn–Al alloys [16,36]. Cu is one of the main alloying additives for Zn–Al alloys to increase hardness, strength and wear resistance [15,26,27,29,33], however, when the Cu content exceeds 2 wt% in Zn–Al alloys, a metastable copper-rich phase (ε) forms causing dimensional instability at elevated temperatures [15,23,25,34,35].

Commercial Zn alloys possess only limited long-term mechanical stability and strength, however, the mechanisms causing this mechanical instability are not yet fully understood. Deeper understanding of the microstructures and the deformation and work hardening mechanisms of Zn alloys are essential for future improvements of their properties and wider application of Zn alloys. The commercial alloy Z410, ZnAl4Cu1 (wt%), (known as Zamak 5 in the US) is a widely used commercial zinc die casting alloy and provides an acceptable combination of strength, ductility as well as excellent plating and finishing characteristics [14,26]. However, there is still a demand to improve its mechanical properties, in particular the mechanical stability. Although the addition of Mg was shown to have beneficial effects in other Zn–Al alloys [16,25,27,32,36], the influence of Mg addition on the mechanical properties and microstructure of ZnAl4Cu1 have not yet been studied. Therefore, it is of great interest to better understand the relation between microstructure and mechanical properties and the underlying deformation and work hardening mechanisms in ZnAlCu alloys and, further, to explore the potential of Mg as a potentially strengthening alloying element in addition to Cu. To this end, we have investigated the microstructure, mechanical properties and deformation behaviour of ZnAl4Cu1 with and without dilute Mg additions at different temperatures and strain rates.

2. Experimental procedures

As a benchmark, plates of commercial zinc alloy Z410 (which will be denoted dc-Z410 in the following) with dimensions 3 mm × 30 mm × 150 mm were produced by hot chamber die casting at HTW Aalen (Beethovenstrasse 1, 73430 Aalen, Germany). On the basis of the composition of alloy dc-Z410, three laboratory alloys gc-Z410, gc-Z410-2 (0.21 wt% Mg) and gc-Z410-3 (0.31 wt% Mg) were prepared in-house by gravity casting. The alloys were melted in air in a resistance furnace using Zn–3.8Al–1.4Cu–0.04Mg (wt%), Zn–3.8Al–0.04Mg (wt%) master alloys, 99.99% pure Al and 99.96% pure Mg. The melt was cast into a preheated copper mould and cooled down in air. The chemical compositions of the alloys were determined by both optical emission spectroscopy using a MAXx spectrometer and inductively coupled plasma atomic emission spectroscopy using a spectro Ciros Vision spectrometer. Optical emission spectroscopy was used to determine the composition of all elements and the accurate composition of Cu and Mg were determined by inductively coupled plasma atomic emission spectroscopy, as listed in Table 1. It was shown by Mohamed et al. [48–50] and Uesugi et al. [51] that

Table 1
Chemical composition of the investigated zinc alloys.

Alloy	Element content, wt%			
	Al	Cu	Mg	Zn
dc-Z410	4.32	0.55	0.04	Bal.
gc-Z410	4.30	0.59	0.04	Bal.
gc-Z410-2	4.30	0.58	0.21	Bal.
gc-Z410-3	4.30	0.59	0.31	Bal.

Fe and Si impurities can significantly influence the mechanical properties of Zn alloys. Therefore, the Fe and Si concentrations were controlled to not exceed 10 ppm Si (5 ppm gc-Z410; 6 ppm gc-Z410-2; 7 ppm gc-Z410-3) and 50 ppm Fe (48 ppm gc-Z410; 41 ppm gc-Z410-2; 43 ppm gc-Z410-3).

Metallographic specimens were prepared by mechanical grinding and polishing to 1/4 μm diamond finish. An intermediate etching step with 90% ethanol + 10% nitrate acid solution to remove the deformation layer was performed between 6 μm and 3 μm diamond polishing steps. The samples were examined with a LEO1530, a JEOL JSM7000F and an FEI Helios Nanolab 600i scanning electron microscope (SEM) at an accelerating voltage of 20 kV using secondary electron microscopy (SE) and backscattered electron microscopy (BSE), energy dispersive X-ray spectroscopy (EDS) and electron backscatter diffraction (EBSD).

The mechanical properties of the as-cast zinc alloys gc-Z410, gc-Z410-2 and gc-Z410-3 were determined by tensile tests at room temperature (RT) ($0.45T/T_m$), 55 °C ($0.5T/T_m$) and 85 °C ($0.55T/T_m$) at strain rates ($\dot{\varepsilon}$) of 5×10^{-4}, 5×10^{-5} and 6×10^{-6} s^{-1}. For comparison, the commercial alloy dc-Z410 was tested at the lowest and highest temperatures and strain rates, i.e. RT and 85 °C ($0.55T/T_m$), $\dot{\varepsilon}$ of 5×10^{-4} and 6×10^{-6}. An electromechanical testing machine (DZM) with an accuracy of 0.17 MPa equipped with an electric furnace was used for these experiments.

In-situ straining experiments were performed in a JEOL 820 SEM with a commercial tensile/compression stage (Kammrath & Weiss Co.) and an integrated resistance furnace. Alloy gc-Z410-3 was tested at RT with a cross head velocity of 0.1 μm s^{-1}, and 85 °C with a cross head velocity of 0.25 μm s^{-1}. The specimens for in-situ straining were standard flat dog-bone shaped samples where the centre region of the gauge length was cut to 0.5 × 0.5 mm^2 (width × length) for observation in the SEM. Due to this deviation from standard tensile specimen dimensions and the resulting inaccuracies in accurately determining the strain within the gauge length, displacements and displacement rates rather than strains and strain rate are given throughout.

3. Results and discussion

3.1. As-cast microstructure

Fig. 1 shows the microstructures of the as-cast laboratory alloys gc-Z410, gc-Z410-2 and gc-Z410-3 and the die-cast commercial alloy dc-Z410 for comparison. It can be seen from Fig. 1a that the microstructure of alloy dc-Z410 is mainly comprised of bright globular grains with an average grain size of 5–6 μm, a fine lamellar and a coarse cellular structure. On the other hand, the microstructure of alloy gc-Z410 (Fig. 1b), which has the same nominal chemical composition as alloy dc-Z410 but was solidified with a slower cooling rate, is comprised of a much coarser dendritic phase with an average grain size of 27–70 μm, a lamellar eutectic structure and some darker areas at the boundaries of the dendrites exhibiting a fine lamellar structure, see the enlarged micrograph in Fig. 1b. Similarly, the microstructures of alloys gc-

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