Effects of Y Additions on the Microstructures and Mechanical Behaviours of as Cast Mg–xY–0.5Zr Alloys

Sihang You, Yuanding Huang, Hajo Dieringa, Emad Maawad, Weimin Gan, Yaping Zhang, Karl Ulrich Kainer, Regine Willumeit-Römer, and Norbert Hort

Previous investigations demonstrated that rare-earth elements (REs) could improve their creep properties effectively. Herein, the influence of Y content on the creep properties of magnesium is investigated systematically with different amount of Y additions. The mechanisms responsible for creep deformation are clarified by the analysis of stress exponent and microstructural characterizations. It is found that the addition of Y in Mg can improve both the ambient strength and high temperature strength owing to its effective solid solution strengthening. At room temperature, the yield strength of Mg–Y alloys has a linear relation with the content of Y. When tested at high temperatures, the yield strength reduces. Compared with pure magnesium, Mg–Y alloys exhibit a high thermal stability even above 200 °C. Small amount of Y addition can improve the creep resistance of Mg largely. With further increasing its content, its contribution to the improvement of creep resistance is weakened for Mg. Under the applied stresses 60–100 MPa and at temperatures of 200–250 °C, the responsible creep mechanism is dislocation controlled. During creep deformation, the Y segregation regions play an important role in hindering the movement of dislocations.

1. Introduction

Magnesium (Mg) alloys have attracted increased interests from industries where weight reduction is concerned, such as automotive and aerospace industries, due to their low density and high specific strength. The most widely used commercial Mg alloys are based on the Mg–Al system, such as AZ series (Mg–Al–Zn) and AM series (Mg–Al–Mn). These systems provide good castability, reasonable room temperature mechanical properties, and good corrosion resistance. However, these conventional Mg alloys have a relatively poor creep resistance at elevated temperatures. This is one of the key issues that limits Mg alloy applications. In the last decades, great efforts have been made to develop Mg alloys with high strength and high creep resistance with the assistance of microstructural design. Rare-earth element (RE) additions, in particular Y, were considered to play an important role in enhancing the creep resistance of Mg alloys. Gravity cast Mg–RE-based alloys were widely reported to have a relatively higher strength and creep resistance at elevated temperatures in comparison to other Mg alloys. The improved creep properties of Mg–RE-based alloys are mainly attributed to the strengthening effect of RE solid solutions and/or RE containing metastable coherent precipitates, such as Gd containing magnesium alloys in which the formed β′ phase has an effective precipitate strengthening. Suzuki et al. reported that the minimum creep rate of solid solution treated Mg–0.2 at%Y was one order of magnitude lower than that of solid solution treated Mg–3 at%Al alloy. Zhu et al. found that, for Mg–2.5RE–0.6Zn alloys, the Nd-rich alloy displayed significantly lower primary creep than the La-rich alloy, owing to the formation of additional Nd-rich precipitates during creep. However, in order to achieve an excellent precipitation hardening response, it always requires...
more than 10 wt% RE content for binary Mg–RE alloys.[19] Commercial Mg–RE alloys such as WE54, WE43 with excellent creep properties also contain a large amount of RE elements. This would lead to prohibitively high costs and make Mg–RE–based alloys suitable only for some extraordinary applications. Consequently, in the last decade, scientists have been searching options to improve the creep resistance of Mg using alloying with non-RE elements and with the development of composites reinforced by ceramic particles such as SiC.[15,20]

In order to develop new creep-resistant magnesium alloys with low contents of REs, it is very necessary to understand their strengthening mechanisms in magnesium during creep deformation. As for the Y element, Suzuki et al. investigated its role in the creep deformation of Mg–Y alloys with its contents in the range from 0.2 to 2.4 mol% Y (about 0.7–9.1 wt%) at relatively higher temperatures above 250 °C (277–377 °C).[17] They found that the creep resistance of magnesium is significantly improved by the addition of yttrium even in a dilute concentration range. The mechanisms responsible for creep deformation are connected with dislocation climbing or dislocation gliding. Unfortunately, they did not examine the creep behavior of Mg–Y alloys in the lower temperature range from 150 to 270 °C. Above 150 °C, as well known, the creep properties of most currently used Mg–Al alloys are sharply deteriorated.[21–23] It is very necessary and important to search effective elements to improve their creep resistance at such temperatures above 150 °C through alloying. The present work thus focused on the investigations on the creep deformation of Mg–Y alloys in such a temperature range from 200 to 250 °C. The responsible mechanisms were explored by microstructural characterization and creep data analysis.

2. Results

2.1. Microstructure Characterizations

Figure 1 presents the optical microstructure of as-cast Mg–Y–Zr alloys. The optical images of all alloys are taken from the same position in their ingots. An equiaxed grain structure is observed in all the Mg–Y–Zr alloys. The grain size first decreases by 0.5 wt % Y addition and then increases after further addition of Y (Table 1). The as-cast WK00 alloy exhibits the finest grains with a size of 84 μm but the most inhomogeneous distribution. However, when adding more Y, the grain distribution become more homogeneous.

Figure 2 shows the scanning electron microscope (SEM) (backscatter electron [BSE]) microstructure of as-cast Mg–Y–Zr alloys. All Mg–Y–Zr alloys consist of α-Mg matrix, Zr-rich particles (bright points shown in the images) and segregation areas with a high concentration of Y alloying element as indicated by arrows. The segregations are located at the interdendritic regions and grain boundaries. In the as-cast WK50 alloy, apart from the Y solute segregation, some bright particles with Y-enriched distribute along dendritic and grain boundaries. The Y content in segregation areas and the volume fraction of intermetallic phases in the as-cast Mg–Y–Zr alloys are indicated in Figure 3. The amount of Y in the segregation areas increases linearly with increasing the Y content. For the as-cast WK50 alloy, the concentration of Y solutes in segregation areas reaches to 4.75 at%, which is close to the maximum Y solubility in α-Mg shown in ref. [24]. When the Y content increase from 0 to 3 wt% the volume fraction of intermetallic phase increases slowly, but jumps to 0.72% at 5 wt% Y.

Figure 4 presents the diffraction patterns of as-cast Mg–Y–Zr alloys using high-energy synchrotron diffraction. These line profiles indicate that, when the Y content is less than 3 wt%, all the Y solutes dissolve in the Mg matrix and the single-phase solid solution is formed. When the Y content is higher than 3 wt%, the peaks corresponding to phase Mg24Y5 appear. Such diffraction results agree with the SEM observation results that the Y-rich particles are also determined to be Mg24Y5 phase.

2.2. Mechanical Properties

2.2.1. Room Temperature

The room temperature mechanical properties of as-cast Mg–Y–Zr alloys are presented in Figure 5. There is a significant increase in both the tensile and compressive strengths with increasing Y addition. It seems that both the tensile yield stress (TYS) and compressive yield stress (CYS) have a linear relation with the content of Y. In contrast, the ultimate tensile stress (UTS) does not linearly increase with the content of Y. When Y content is less than 3 wt%, the tensile elongation to failure increases remarkably with increasing alloy addition. However, with further increasing the content of Y from 3 to 5 wt%, the tensile elongation drops from 15% to 10%. Similar to the evolution of tensile elongation, the compressibility also firstly increases to 33% at 0.5 wt% of Y addition and then decreases with increasing Y content.

2.2.2. Elevated Temperatures

Compressive Properties: Considering the present creep deformation was carried out in a compressive way, the compressive properties of Mg–Y–Zr alloys at high temperatures were also measured (Figure 6). With increasing the test temperature, the CYS of all alloys decreases and the compressibility increases. Interestingly, the reduction in CYS and the increment in compressibility of as-cast Mg-0.5Zr alloy are much greater and faster than that of the alloys containing Y, especially when the test temperature is more than 200 °C. Above 200 °C, the CYS of alloys containing Y still keeps quite stable, indicating that the addition of Y significantly improves the thermal stability of as-cast Mg alloy.

Creep Properties: Figure 7 and 8 only present the creep strain curves and creep rate curves of as-cast Mg–Y–Zr alloys at 225 °C, 80 MPa, respectively. The creep properties at other temperatures are presented later to discuss the responsible creep mechanisms (in the section of Discussion). As shown in Figure 7 and 8, the increasing Y content up to 5 wt% results in a significant improvement of creep properties. All Y containing alloys exhibit a similar creep behavior, i.e., primary, secondary and tertiary stages. Regarding different alloys, they have different durations of secondary stage. As shown in Figure 8, when the content of Y is more than 3 wt%, the creep rate decreases largely.
Microstructure After Creep Deformation

The alloy with the highest content of Y was selected to analyze the microstructure after creep deformation. Figure 9 shows the bright field (BF) transmission electron microscope (TEM) microstructure of as-cast WK50 alloy crept at 225 °C and 80 MPa for 200 h. A localized strain region was found closed to the grain boundary (Figure 9b) in which some dislocation tangles were observed. In this region, the information about precipitates was not indicated by the selected area electron diffraction (SAED) analysis, demonstrating that the dark contrast is more likely caused by dislocations rather than the precipitates. In addition, fine particles were observed at grain boundaries in this alloy with 5 wt% Y (see the marks in Figure 9a). They were identified as the phase Mg24Y5. This agrees with results obtained by high-energy synchrotron diffraction.

3. Discussion

3.1. Microstructure

Due to the high cooling rate associated with direct chill casting, the solidification of as-cast Mg–Y–Zr alloys proceeds under a nonequilibrium condition. The alloys tend to form a supersaturated single-phase microstructure. Meanwhile, the distribution...
Figure 2. SEM (BSE) micrographs of as-cast Mg–Y–Zr alloys: a) Mg, K0; b) WK00; c) WK10; d) WK30; and e) WK50.

Figure 3. Y content in segregation areas and volume fraction of intermetallic phases as a function of Y content in the as-cast Mg–Y–Zr alloys.

Figure 4. Synchrotron diffraction line profiles of as-cast Mg–Y–Zr alloys.
Figure 5. Room temperature mechanical properties of as-cast Mg–Y–Zr alloys, a) tensile properties and b) compressive properties.

Figure 6. Compressive properties of as-cast Mg–Y–Zr alloys as a function of test temperature: a) compressive yield strength and b) compressibility.

Figure 7. Creep strain curves for the as-cast Mg–Y–Zr alloys tested at 225 °C, 80 MPa.

Figure 8. Creep strain rates for the as-cast Mg–Y–Zr alloys tested at 225 °C, 80 MPa.
of solute Y is not homogeneous, and it is segregated at the dendiric and grain boundaries. The alloying element Y has a high solid solubility in Mg. During nonequilibrium solidification, a large amount of Y solutes could segregate along the grain boundaries because some of the solutes distribute in the Mg matrix. In the alloys containing less than 3 wt% of Y no Mg–Y compounds were detected. Only in the as-cast alloy with 5 wt% Y a small amount of Mg–Y intermetallic particles was observed along grain boundaries. According to the results of SEM, synchrotron diffraction and TEM, these particles were confirmed to be Mg24Y5, which agrees with that indicated by Mg–Y binary phase diagram.

Zr is an extremely effective grain refiner for Mg alloys that contain no Al, Mn, Si, and Fe. The undissolved Zr particles could act as potential sites for heterogeneous nucleation and the dissolved Zr with high growth restriction factor value allows to rapidly build up an effective constitutional supercooling zone ahead of the growing crystal, resulting in the potential grain refinement. Therefore, the grain structure of all the as-cast alloys is relatively homogeneous and the grain sizes are not very coarse due to the addition of Zr. Besides the role of Zr, the grain size is also affected by the addition of Y. It was reported that the addition of Y had a grain refinement effect on the as-cast Mg–Y alloys. However, in this work, the grain size of as-cast Mg–Y–0.5Zr alloys first remarkably decreases at 0.5 wt% Y, and then slightly increases after adding more Y up to 5 wt%. This is mainly due to the high cooling rate during casting and the addition of Zr, which makes the grain refinement effect of Y negligible. The growth restriction factor Q value was widely used to describe the impact of alloying elements on the grain size of as-cast alloys which is related to the developing rate of constitutional supercooling zone at the early stage of grain growth. In a binary system, the value of Q can be calculated by the following equation

$$Q = mC(k - 1)$$

where m is the slope of the liquidus line, k is the distribution coefficient and C is the composition of the solute in a binary alloy. Higher Q value stands for a faster constitutional supercooling zone development and thus, a better grain refinement effect of an alloying element. At C = 1% of alloying element, the growth restriction factor Q of Y (m = −3.4, k = 0.5) was calculated as 1.70, which is smaller than that for other alloying elements such as Ca (m = −12.67, k = 0.06) with a value of 11.94. It indicates that Y addition has a weaker effect on the grain refinement of Mg.

### 3.2. Tensile and Compressive Properties at Ambient and High Temperatures

After the addition of Y, both the room temperature and high temperature strength of Mg were improved (Figure 5 and 6). Moreover, the thermal stability of Mg is also increased at high temperatures, especially above 200 °C. Such strength improvements are caused by the solid solution strengthening of Mg. In the investigated alloys, the amount of intermetallics is less (Figure 2 and 3). The contribution to strength from particle strengthening could be ignored. Besides the solid solution strengthening of Y, the improvement of ambient strength is also partly attributed to the reduction in grain size after Y addition (Figure 1). According to Hall–Petch principle, small grains are helpful for the improvements of not only strength but also elongation. Regarding the solid solution strengthening of Y in Mg matrix, it will be further discussed in detail in the following section.

Compared with Mg, Mg–Y alloys have a higher ambient elongation (Figure 5). As aforementioned, apparently one of main mechanisms is attributed to smaller grains caused by Y addition. However, another important factor should not be neglected. Many previous investigations demonstrated when adding those RE elements such as Gd or Y with a high solid solubility into Mg they can cause a decline in the ratio of lattice parameter c to a. Such reduction in c value then leads to the increment of critical resolved shear stress for the basal slip, and is therefore beneficial for the activations of nonbasal slip like pyramidal slip. The homogeneity of deformation is improved and then the elongation increases. When adding 5 wt% Y to Mg, the elongation of
Mg–5Y–Zr alloy reduces. This should be caused by its grain coarsening. With increasing Y content, the grain size increases (Figure 1). In addition, the existence of some large second phase particles in this alloy may additionally result in its reduction in the elongation, although it may not be the dominant mechanism. Near such large particles, owing to the incompatible deformation between them and Mg matrix, the cracks could easily initiate at their interface.

The increment of the test temperature favors the activation of dislocation slip and the occurrence of recovery, resulting in the reduction of yield stress and the increasing of elongation. It was reported that nonbasal slips in Mg can be activated when the test temperature is above 200 °C.[38–40] As a result, the yield stress of Mg largely decreases when tested at the temperatures of more than 200 °C. In contrast, owing to the solid solution strengthening of Y (discussed later in detail), the yield stress of Mg–Y alloys was kept stable at the testing temperature 200 °C.

### 3.3. Creep Deformation

#### 3.3.1. Creep Mechanisms

The dependence of minimum creep rate \( \dot{\varepsilon} \) on the applied stress \( \sigma \) and the temperature in Kelvin is given as follows:[41,42]

\[
\dot{\varepsilon} = A\sigma^n \exp\left(-\frac{Q_c}{RT}\right)
\]

where \( A \) is a constant related to the frequency factor and Burger’s vector, \( \sigma \) is the applied creep stress, \( Q_c \) is the activation energy of creep, \( R \) is a universal gas constant (8.314 J mol\(^{-1}\) K\(^{-1}\)), and \( n \) is the stress exponent. Equation (2) is rewritten by taking logarithms on both sides of the Equation (2)

\[
\ln\dot{\varepsilon} = \ln A + n \ln \sigma - \frac{Q_c}{RT}
\]

When the creep temperature is fixed, the stress exponent \( n \) can be determined by the slope of \( \ln\dot{\varepsilon} \) against \( \ln\sigma \). Previous investigations claimed that the stress exponent \( n = 3 \) is related to viscous glide of dislocations, \( n = 5 \) to dislocation climbing at high temperatures, and \( n = 7 \) to dislocation climbing at low temperatures.[14,21,43–46]

Figure 10 shows the minimum creep rate as a function of stress with stress exponents (\( n \) values) for the as-cast Mg–Y–Zr alloys at: a) 200 °C and b) 250 °C.

![Figure 10. Minimum creep rate as a function of stress with stress exponents (\( n \) values) for the as-cast Mg–Y–Zr alloys at: a) 200 °C and b) 250 °C.](image)

The activation energy \( Q \) values at 80 MPa are around 130 kJ mol\(^{-1}\) for the alloys Mg and WK00 and about 190 kJ mol\(^{-1}\) for that with Y content more than 1 wt% (Figure 11). The Y content does not affect the activation energy significantly. The reported activation energy \( Q \) for the self-
diffusion of Mg varies from 81 to 135 kJ mol\(^{-1}\), and \(Q\) for the cross-slip in precipitation hardened alloys is from 229 to 325 kJ mol\(^{-1}\).[8,11] The \(Q\) values obtained in the present work suggest that dislocation climbing plays a dominant role during creep deformation in the temperature range from 200 to 250 °C. This conclusion agrees with previous results for other Mg–RE alloys.[10] The present dislocation controlling creep mechanism concluded from \(n\) and \(Q\) values is also supported by TEM observations on the microstructure of sample crept at 225 °C and 80 MPa (Figure 9).

3.3.2. Effects of Y

After the addition of Y up to 5 wt%, the creep properties of Mg are apparently improved at both the low and high temperatures under a stress range of 60–100 MPa. Based on the microstructural observations, it can be concluded that such creep property improvement is closely related to the content of Y solutes in solid solution of Mg, i.e., to the solid solution strengthening of Y. First, although the addition of Y reduces the grain size of Mg, its content has little effect on the grain size of Mg–Y alloys (Table 1 and Figure 1). As known, smaller grains could deteriorate the creep resistance. If only considering the factor of grain size caused by Y addition, the creep resistance of Mg–Y–Zr alloys should reduce. However, compared with Mg–Zr alloy, all Mg–Y–Zr alloys have a higher creep resistance (Figure 7). Second, the intermetallic compounds were not obviously observed after the addition of Y (Figure 2 and 3). Only a slightly increase in the volume fraction of intermetallic compounds Mg\(_2\)Y\(_3\) was found when the Y content increases to 5 wt%. Their formation and existence could contribute to the improvement of creep resistance through particle strengthening or/and pinning effects of grain boundaries. However, microstructural observations demonstrated that the amount of Mg\(_2\)Y\(_3\) is very small. It is convincing that such small amount of intermetallics is not enough to explain the strongly increased creep properties with increasing Y content. Moreover, present TEM observations on the crept Mg–0.5Y–0.5Zr sample demonstrated that no dynamic precipitation occurred during creep deformation. Consequently, the improvement of creep resistance owing to the particle strengthening could be neglected. Finally, the creep resistance of Mg–Y–Zr alloy is proportional to the content of Y. As shown in Figure 3, the Y content in segregation areas significantly increases with increasing Y content. Microstructural observations indicated that such Y segregation regions located close to grain boundaries. During creep deformation, dislocations could act as nucleation points for the clustering of Y solutes. Then, the Y solute clusters furtherly hinder the dislocations motion and result in the dislocation tangles in these regions. Thus, the localized strain regions close to grain boundaries were formed. The higher density of dislocations existed in these regions indicates that the segregation of Y solutes along dendritic and grain boundaries plays a very important role on hindering the dislocation motion during creep deformation (Figure 9).

Although the effects of RE segregation on creep were rarely mentioned previously, in fact, that RE saturated solid solution plays an important role in improving the creep resistance of Mg alloys was widely reported.[8,11,16] Zhu et al. compared the effects of several RE elements (Nd, Ce and La) on the creep properties of die-cast Mg alloys. They found that the best creep property was obtained for Mg–Nd alloy.[48] It was suggested that the creep resistance was highly related to the solid solubility of RE element in Mg matrix rather than dependent on the grain boundary pinning effect by their intermetallic compounds. Furthermore, Gavras et al. considered that the solid solution strengthening of RE solutes played an even more important role than dynamic precipitation strengthening in high pressure die casting Mg–RE alloys.[49] It was found that the T4 heat-treated Mg–La–Y alloy exhibited a lower secondary creep rate than Mg–La–Nd alloy even though the dynamic precipitation occurred in the latter alloy.[49] In the current work, the minimum creep rate of as-cast alloys reduces sharply even by the addition of small amount of Y owing to its effective solid solution strengthening. However, it should be noticed that the improvement of creep resistance by solid solution strengthening of Y element is mainly dependent on its solid solubility in Mg matrix. As shown in Figure 7, with increasing the content of Y, its strengthening efficiency is weakened. A similar phenomenon was also reported in the creep behavior of die-cast Mg–La–RE alloys.[49] It was found that further increment in ternary RE content did little to improve the creep properties after a critical content of RE is reached.

4. Conclusions

The influence of Y content on the creep resistance of Mg–Zr alloys was investigated by the characterizations of creep deformation and microstructural observations. Following conclusions can be obtained: 1) After the addition of 0.5 wt% Y to Mg–0.5Zr, the grain size reduces. The solute Y goes into solid solution in Mg and is enriched in the segregation regions near the dendritic and grain boundaries. With increasing the amount of Y addition, the content of Y in these segregation regions also increases. Only very small amount of intermetallic particles Mg\(_2\)Y\(_3\) is formed at dendritic and grain boundaries in Mg–Y–Zr alloy with the addition of 5 wt% Y. 2) Both the room temperature tensile and compressive yield stress linearly increase with increasing the amount of Y addition. The transient plastic deformation of Mg–Y–Zr alloys monotonously increases when the content of Y increases to 3 wt%. With further increasing the content of Y, it then reduces. Increasing the test temperature reduces the yield stress and increases the compressibility. Y addition is beneficial for improving the thermal stability of Mg at high temperatures. 3) Small addition of Y can improve the creep resistance of Mg largely. With increasing the content of Y, its contribution to the improvement of creep resistance is weakened. Under the applied stresses of 40–100 MPa and at temperatures of 200 to 250 °C, the responsible creep mechanism is the dislocation controlled. During creep deformation, the Y segregation regions play an important role in hindering the movement of dislocations.

5. Experimental Section

Five alloys with different Y concentrations (wt%), Mg–0.5Zr, Mg–0.5Y–0.5Zr, Mg–1.0Y–0.5Zr, Mg–3.0Y–0.5Zr, and Mg–5.0Y–0.5Zr (Table 2), were prepared by permanent mould direct chill casting.[50] Pure Mg was melted in a mild steel crucible under a protective atmosphere (Ar + 2
Table 2. Nominal chemical compositions of experimental alloys (wt%).

| Alloys  | Compositions | Y  | Zr | Mg |
|---------|--------------|----|----|----|
| Mg (K0) | Mg–0.5Zr     | –  | 0.5| Bal.|
| WK00    | Mg–0.5Y–0.5Zr| 0.5| 0.5| Bal.|
| WK10    | Mg–1Y–0.5Zr  | 1  | 0.5| Bal.|
| WK30    | Mg–3Y–0.5Zr  | 3  | 0.5| Bal.|
| WK50    | Mg–5Y–0.5Zr  | 5  | 0.5| Bal.|

vol% SF3. Mg–33.3 wt% Zr master alloy were then added into the melt at 750 °C. The melt was then homogenized by mechanical stirring under 200 rpm for 20 min. After that, the melt was poured into a steel crucible preheated to 680 °C and held at 680 °C for 20 min with gas protection. Finally, the melt was solidified by lowering the crucible into cooling water with a rate of 10 mm s⁻¹.

The specimens for optical microscopy (OM) were ground with silicon carbide abrasive paper, polished with water-free colloidal silica (OPS) and then etched with acetic-picric solution (30 mL deionized water, 140 mL ethanol, 7 mL glacial acetic acid, and 8 g picric acid). OM observations were performed with a Leica DM1500 microscope with a digital camera attached. The average grain size was measured by the linear intercept method from the micrographs. Microstructures were further investigated using a TESCAN Vega SB-U III SEM under an accelerating voltage of 15 kV and working distance of 15 mm in BSE mode. The concentration of Y solutions in segregation areas was measured by EDS analysis. More than 30 positions in segregation areas were selected randomly for EDS analysis, and the average value and errors of Y concentration (at%) were calculated as the final results. The volume fraction of the intermetallic phases was determined by SEM BSE image analysis using ImageJ software. The threshold and contrast were adjusted to obtain a high contrast between the intermetallic phase regions and the matrix. At least 10 micrographs were used to determine the amount of these phases. Specimens for TEM were first ground mechanically to a thickness of about 120 μm and then thinned by twin jet electropolishing using a polishing solution of 2.5% HClO₄ and 97.5% ethanol at −45 °C and a voltage of 40 V. The TEM examinations were carried out on a Philips CM200 instrument operating at 200 kV. The bright-field mode was used for imaging the microstructure. The diffraction patterns of intermetallic particles were obtained through SAED. Additionally, synchrotron radiation diffraction was used for the phase analysis, and was carried out at the high-energy materials science beamline (HEMS, P07B) of the Helmholtz-Zentrum Hereon at PETRA III of DESY (Deutsches Elektronen-Synchrotron). A monochromatic beam was used with a wavelength of 0.014235 nm and a beam cross-section of 0.7 × 0.7 mm². The acquisition time of each diffraction pattern was set to 0.5 s. The diffraction patterns (Debye-Scherrer rings) were recorded by a detector with an effective pixel size of 200 μm, placed 1950 mm behind the sample, calibrated by a LaB₆ standard powder sample. The diffraction patterns were integrated and analyzed with the software Fit2D.

Tensile and compressive tests were performed using a Zwick 050 testing machine (Zwick GmbH & Co., KG, Ulm, Germany). The tensile specimens have a gauge length of 30 mm, a diameter of 6 mm and the compression specimens were cylinders with a dimension of 11 × 16.5 mm² (diameter × height). Both the tension and compression tests were performed under a strain rate of 1 × 10⁻⁴ s⁻¹. At least three samples were tested for each condition. All the tension tests were performed at room temperature. Compression tests were carried out not only at room temperature but also at high temperatures 175, 200, 225, and 250 °C, respectively. At high temperatures, the samples were placed into the testing machine after the chamber reaching the testing temperature. They were then held for around 10 min at the testing temperature before starting the tests. The cylindrical specimens of 15 mm in length and 6 mm in diameter were used for the compression creep tests. The constant stress compressive creep tests were performed for the as-cast Mg–Y–Zr alloys at 200, 225, and 250 °C under a range of stresses (40–100 MPa) using Applied Test System (ATS) creep machines.

Acknowledgements

The authors acknowledge Mr. G. Meister for preparing the alloys. S.Y. gratefully thanks the China Scholarship Council (201506120049) for the award of a fellowship and funding. The authors acknowledge DESY (Hamburg, Germany), a member of the Helmholtz Association HGF, for the provision of experimental facilities. Parts of this research were carried out at the P07 beamline.

Open access funding enabled and organized by Projekt DEAL.

Conflict of Interest

The authors declare no conflict of interest.

Data Availability Statement

The data that support the findings of this study are available from the corresponding author upon reasonable request.

Keywords

creep deformation, magnesium alloys, mechanical properties, microstructures

Received: August 10, 2021
Revised: December 12, 2021
Published online: January 18, 2022

[1] K. U. Kainer, Wiley-VCH GmbH, Weinheim, Germany 2003.
[2] K. U. Kainer, Y. D. Huang, H. Dieringa, N. Hort, Mater. Sci. Forum 2009, 638–642, 73.
[3] A. A. Luo, J. Magn. Alloys 2013, 1, 2.
[4] A. A. Luo, Int. Mater. Rev. 2004, 49, 13.
[5] N. Mo, Q. Tan, M. Bermingham, Y. Huang, H. Dieringa, N. Hort, M.-X. Zhang, Mater. Des. 2018, 135, 422.
[6] N. Hort, Y. Huang, K. Kainer, Adv. Eng. Mater. 2006, 8, 235.
[7] N. Stanford, R. Cottam, B. Davis, J. Robson, Acta Mater. 2014, 78, 1.
[8] L. Gao, R. S. Chen, E. H. Han, J. Alloys Compd. 2009, 472, 234.
[9] L. Gao, R. S. Chen, E. H. Han, J. Alloys Compd. 2009, 481, 379.
[10] H. Hu, A. Yu, N. Y. Li, J. E. Allison, Mater. Manuf. Process. 2003, 18, 687.
[11] B. L. Mordike, Mater. Sci. Eng. A 2002, 324, 103.
[12] M. O. Pekgulyuz, A. A. Kaya, Adv. Eng. Mater. 2003, 5, 866.
[13] S. Zhu, M. Easton, T. Abbott, J.-F. Nie, M. Dargusch, N. Hort, M. Gibson, Metall. Mater. Trans. A 2015, 46, 3543.
[14] S. M. Zhu, J. F. Nie, M. A. Gibson, M. A. Easton, P. Bakke, Metall. Mater. Trans. A 2012, 43, 4137.
[15] C. Xu, T. Nakata, G. H. Fan, X. W. Li, C. Z. Tang, S. Kamado, J. Magn. Alloys 2019, 7, 388.
[16] M. Suzuki, H. Sato, K. Maruyama, H. Oikawa, Mater. Sci. Eng. A 1998, 252, 248.
[17] M. Suzuki, H. Sato, K. Maruyama, H. Oikawa, Mater. Sci. Eng. A 2001, 319–321, 751.
[18] S. M. Zhu, M. A. Gibson, J. F. Nie, M. A. Easton, G. L. Dunlop, Metall. Mater. Trans. A 2009, 40, 2036.
[19] N. Hort, Y. Huang, D. Fechner, M. Störmer, C. Blawert, F. Witte, C. Vogt, H. Drücker, R. Willumeit, K. U. Kainer, F. Feyerabend, Acta Biomater. 2010, 6, 1714.
[20] P. Zhou, S. Zhang, M. Li, H. Wang, W. Cheng, L. Wang, H. Li, W. Liang, Y. Liu, J. Magn. Alloys 2020, 8, 944.
[21] H. Dieringa, Y. D. Huang, P. Maier, N. Hort, K. U. Kainer, *Mater. Sci. Eng. A Struct.* 2005, 410, 85.

[22] B. Nami, H. Razavi, S. Mirdamadi, S. G. Shabestari, S. M. Miresmaeili, *Metall. Mater. Trans. A* 2010, 41, 1973.

[23] A. Srinivasan, J. Swaminathan, M. K. Gunjan, U. T. S. Pillai, B. C. Pai, *Mater. Sci. Eng. A* 2010, 527, 1395.

[24] A. A. Nayeb-Hashemi, J. B. Clark, *Phase Diagram of Binary Magnesium Alloys*, ASM International, Metal Park, OH 1988.

[25] Z. Wang, Y. D. Huang, A. Srinivasan, Z. Liu, F. Beckmann, K. U. Kainer, N. Hort, *Mater. Des.* 2013, 47, 90.

[26] D. H. Stjohn, M. A. Easton, M. Qian, J. A. Taylor, *Metall. Mater. Trans. A* 2013, 44A, 2935.

[27] D. H. StJohn, M. Qian, M. A. Easton, P. Cao, Z. Hildebrand, *Metall. Mater. Trans. A* 2005, 36A, 1669.

[28] Y. Ali, D. Qiu, B. Jiang, F. Pan, M.-X. Zhang, *J. Alloys Compd.* 2015, 619, 639.

[29] Q. Zhang, Q. A. Li, J. Chen, X. Y. Zhang, *Adv. Mater. Res.* 2013, 750–752, 603.

[30] M. A. Easton, D. H. StJohn, *Acta Mater.* 2001, 49, 1867.

[31] T. E. Quested, A. T. Dinsdale, A. L. Greer, *Acta Mater.* 2005, 53, 1323.

[32] Y. Huang, W. Gan, K. U. Kainer, N. Hort, *J. Magn. Alloys* 2014, 2, 1.

[33] A. Kula, K. Jia, R. K. Mishra, M. Niewczas, *Metall. Mater. Trans. B* 2016, 47, 3333.

[34] S. Miura, S. Imagawa, T. Toyoda, K. Ohkubo, T. Mohri, *Mater. Trans.* 2008, 49, 952.

[35] H. K. Lin, D. H. Kim, J. Y. Lee, W. T. Kim, D. H. Kim, *Mater. Lett.* 2008, 62, 2271.

[36] Y. Xu, F. Gensch, Z. Ren, K. U. Kainer, N. Hort, *Progr. Nat. Sci. Mater. Int.* 2018, 28, 724.

[37] Q. Peng, J. Meng, Y. Li, Y. Huang, N. Hort, *Mater. Sci. Eng. A* 2011, 528, 2106.

[38] H. Wang, C. J. Boehlert, Q. Wang, D. Yin, W. Ding, *Metall. Mater. Trans. A* 2016, 47, 2421.

[39] A. Chapuis, J. H. Driver, *Acta Mater.* 2011, 59, 1986.

[40] H. Fan, Q. Wang, X. Tian, J. A. El-Awady, *Scr. Mater.* 2017, 127, 68.

[41] Y. Li, T. G. Langdon, *Scr. Mater.* 1997, 36, 1457.

[42] H. Oikawa, S. Karashima, *Metall. Trans.* 1974, 5, 1179.

[43] H. Dieringa, Y. Huang, P. Wittke, M. Klein, F. Walther, M. Dikovits, C. Poletti, *Mater. Sci. Eng. A* 2013, 585, 430.

[44] Y. Huang, H. Dieringa, K. U. Kainer, N. Hort, *Fatigue Fract. Eng. M* 2013, 36, 308.

[45] J. G. Wang, L. M. Hsiung, T. G. Nieh, M. Mabuchi, *Mater. Sci. Eng. A* 2001, 315, 81.

[46] H. Yang, Y. Huang, B. Song, K. U. Kainer, H. Dieringa, *Mater. Sci. Eng. A* 2019, 755, 18.

[47] S. S. Vagarali, T. G. Langdon, *Acta Metall.* 1981, 29, 1969.

[48] S. M. Zhu, M. A. Gibson, M. A. Easton, J. F. Nie, *Scr. Mater.* 2010, 63, 698.

[49] S. Gavras, S. M. Zhu, J. F. Nie, M. A. Gibson, M. A. Easton, *Mater. Sci. Eng. A* 2016, 675, 65.

[50] F. R. Elsayed, N. Hort, M. A. Salgado-Ordorica, K. Kainer, *Mater. Sci. Forum* 2011, 690, 65.