Microstructure and mechanical properties of LPSO phase dominant $\text{Mg}_{85.8}\text{Y}_{7.1}\text{Zn}_{7.1}$ and $\text{Mg}_{85.8}\text{Y}_{7.1}\text{Ni}_{7.1}$ alloys

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Abstract – $\text{Mg}_{85.8}\text{Y}_{7.1}\text{Zn}_{7.1}$ ($\text{Mg}_{12}\text{Y}_1\text{Zn}_1$) and $\text{Mg}_{85.8}\text{Y}_{7.1}\text{Ni}_{7.1}$ ($\text{Mg}_{12}\text{Y}_1\text{Ni}_1$) alloys composed primarily of long-period stacking ordered (LPSO) phase were prepared by conventional casting and were further processed by hot extrusion or equal-channel angular pressing (ECAP) to investigate the effect of substitution of Zn with Ni on microstructure and mechanical properties. While a mixture of 18R and 14H structures of the LPSO phase was observed in the cast $\text{Mg}_{85.8}\text{Y}_{7.1}\text{Zn}_{7.1}$ alloy after homogenization treatment, the LPSO phase in the $\text{Mg}_{85.8}\text{Y}_{7.1}\text{Ni}_{7.1}$ alloy was found to consist of 18R only. Beside the LPSO phase, both alloys contained small amounts of α -Mg and intermetallic phases, *viz.* MgZn_2Y (W-phase) in the $\text{Mg}_{85.8}\text{Y}_{7.1}\text{Zn}_{7.1}$ alloy and MgNi_2 and MgNi_4Y in the $\text{Mg}_{85.8}\text{Y}_{7.1}\text{Ni}_{7.1}$ alloy. Mechanical testing of the extruded or ECAP processed alloys showed that the substitution of Zn with Ni significantly improved the strength of the $\text{Mg}_{85.8}\text{Y}_{7.1}\text{Zn}_{7.1}$ alloy, but at the expense of ductility. The mechanical properties of the alloys are discussed in light of the microstructural observations.

Keywords: Magnesium alloys; Long-period stacking ordered (LPSO) phase; Mechanical properties; Mg–Y–Zn; Mg–Y–Ni

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1. Introduction

In recent years, ternary Mg–RE–X (where RE stands for rare earth including Y, Gd, Dy, Ho, Er, Tm or Tb; X represents Zn, Cu, Ni or Co) alloys have attracted considerable attention due to their unique microstructure and excellent mechanical properties [1]. Typically, the microstructure of such alloys contains a mixture of α -Mg grains and intermetallic particles that exhibit a range of long-period stacking ordered (LPSO) structures, in which 18R and 14H are the most common ones. A distinctive feature of such LPSO structures is their lamellar morphology that comprises alternating layers of α -Mg and intermetallic blocks. The most investigated systems are Mg–Y–Zn based alloys, which exhibit a wide range of mechanical properties, depending on alloy composition [2-4] and thermomechanical processing applied after casting [5-9].

The accumulated experimental results indicate that mechanical properties (strength and ductility) of the Mg–Y–Zn alloys can be tailored by proper control of the volume fraction and size of lamellae of the LPSO structures. The volume fraction of the LPSO structures can be controlled by the alloy composition. Investigations of the influence of the Y and Zn contents [2] and Y/Zn mole ratio [4] showed a range of mechanical properties, namely 330–400 MPa tensile strength and 11–13% ductility. By utilizing powder rapid solidification, an impressive gain in mechanical properties with yield strength reaching up to 600 MPa and 5% tensile ductility was obtained in Mg₉₇Y₂Zn₁ [5].

As Mg–Y–Zn ternary alloys are not precipitation hardenable [10], the strengthening in these alloys is normally obtained by plastic deformation. Alloying of the Mg–Y system with Ni, however, has been shown to result in an age hardening response. The study of extruded Mg₉₄Y₄Ni₂ alloy [11] revealed the precipitation of fine β' phase in the matrix after solution treatment and aging. A tensile strength of 453 MPa with limited ductility of 2.4 % was reported, which was related to the coexistence of the LPSO phase and β' precipitates [11].

The idea of simultaneous alloying of Mg alloys with Zn and Ni has been discussed over the last decade [12-14]. The addition of 0.5 wt.% Ni to Mg–3wt.% Y–0.5wt.% Zn alloy was shown to decrease the density of stacking faults (SF) while increasing the density of precipitates [12]. More recently, a positive effect of 1 at.% Ni substitution for Zn was reported in Mg₉₄Y₄Zn₂ alloy [14], with the tensile strength increased by 27 MPa (from 390 MPa to 417 MPa), tensile yield strength by 22 MPa (from 246 MPa to 268 MPa) and tensile elongation by 0.6% (from 2.8% to 3.4%). The improvements in mechanical properties were ascribed to the

enhanced solid solution strengthening and thermal stability of the LPSO phase due to substitution of Zn with Ni.

To further explore the effect of substitution of Zn with Ni on the LPSO phase and ultimately the mechanical properties of Mg–Y–Zn system, it is ideal to use alloys composed primarily of the LPSO phase. Based on the previous work [15, 16] which showed that the LPSO phase in Mg–Y–Zn alloys most likely has a stoichiometric composition of $\text{Mg}_{12}\text{Y}_1\text{Zn}_1$, we prepared $\text{Mg}_{85.8}\text{Y}_{7.1}\text{Zn}_{7.1}$ ($\text{Mg}_{12}\text{Y}_1\text{Zn}_1$) and $\text{Mg}_{85.8}\text{Y}_{7.1}\text{Ni}_{7.1}$ ($\text{Mg}_{12}\text{Y}_1\text{Ni}_1$) alloys in an attempt to maximize the LPSO phase. This paper presents detailed microstructural analysis and mechanical evaluation of the aforementioned alloys.

2. Experimental methods

The $\text{Mg}_{85.8}\text{Y}_{7.1}\text{Zn}_{7.1}$ and $\text{Mg}_{85.8}\text{Y}_{7.1}\text{Ni}_{7.1}$ alloys were prepared from high purity (>99.9%) elemental metals by induction melting in a mild steel crucible at $\sim 760^\circ\text{C}$ under an atmosphere of a mixture of HFC-R134a and argon, and casting into a steel mould pre-heated to $\sim 200^\circ\text{C}$. The cast ingots were given a homogenization treatment at 450°C for 24 h, followed by quenching into warm water of about 50°C .

Cylindrical blocks with a diameter of 35 mm for extrusion and of 10 mm for ECAP were machined from the homogenised ingots. The extrusion was carried out at 450°C with a ram speed of 0.5 mm/s and an extrusion ratio of 12.25:1, which introduced an equivalent strain of ~ 2.5 into the extruded rods. Four passes of ECAP (Route B_C in which the specimen is rotated by 90° about the pressing direction between the passes) were performed using a 90° die [8] at 450°C , with a back pressure of 150 MPa and a pressing speed of 1 mm/s. Each ECAP pass imparted an equivalent strain of ~ 1.15 onto the processed rod. Cylindrical specimens with a diameter of 4 mm and a height of 7 mm were machined from the extruded or ECAP processed rods and were compression tested at room temperature, with the compression direction along the extrusion or the ECAP pressing direction. Three specimens were used for each test condition. The crosshead speed was 0.3 mm/min, corresponding to a nominal strain rate of $7.1 \times 10^{-4} \text{ s}^{-1}$.

Scanning electron microscopy (SEM), transmission electron microscopy (TEM), high resolution transmission electron microscopy (HRTEM) and energy dispersive X-ray (EDX) spectroscopy were used to characterise the microstructure in the alloys before and after

extrusion or ECAP. The TEM foils were prepared by low-angle ion milling using a Gatan Precision Ion Polishing System (PIPS) at 4 keV with an incident angle of 4° and were examined in either FEI Tecnai-T20 or FEI Tecnai-F20 electron microscope operated at 200 kV.

3. Results

3.1. Microstructure after homogenisation treatment

The typical SEM images of the cast $\text{Mg}_{85.8}\text{Y}_{7.1}\text{Zn}_{7.1}$ and $\text{Mg}_{85.8}\text{Y}_{7.1}\text{Ni}_{7.1}$ alloys after the homogenisation at 450°C for 24 h are shown in Fig. 1. It is apparent that neither alloy is fully composed of the LPSO phase as expected; however, the microstructures are predominated by LPSO phase with fine lamellar features. For the $\text{Mg}_{85.8}\text{Y}_{7.1}\text{Zn}_{7.1}$ alloy, the microstructure is largely composed of the LPSO phase (grey) and small amounts of another intermetallic phase (bright) and $\alpha\text{-Mg}$ (dark). It appears that the intermetallic particles are mostly present with the $\alpha\text{-Mg}$ phase as a common eutectic, indicating a eutectic reaction in solidification. However, some $\alpha\text{-Mg}$ phase is distributed within the LPSO phase, implying a possible peritectic reaction during solidification. With the substitution of Zn with Ni, the fraction of LPSO phase increases while the fraction of eutectic or intermetallic phase decreases. In addition, the intermetallic phase appears to be fully separated from the $\alpha\text{-Mg}$ phase in the $\text{Mg}_{85.8}\text{Y}_{7.1}\text{Ni}_{7.1}$ alloy as compared with the common eutectic morphology in the $\text{Mg}_{85.8}\text{Y}_{7.1}\text{Zn}_{7.1}$ alloy. Furthermore, there is a significant increase in the size of the intermetallic particles in the $\text{Mg}_{85.8}\text{Y}_{7.1}\text{Ni}_{7.1}$ alloy, most particles being oblong. It is noted that the $\alpha\text{-Mg}$ phase is also present in the $\text{Mg}_{85.8}\text{Y}_{7.1}\text{Ni}_{7.1}$ alloy, but only in a small volume fraction.

The alloys were further characterised by TEM to reveal the structure of the LPSO phase and the identity of the other intermetallic phases observed. Fig. 2 shows TEM micrographs of the LPSO phase in the $\text{Mg}_{85.8}\text{Y}_{7.1}\text{Zn}_{7.1}$ alloy. Based on the lattice fringes and electron diffraction patterns, the LPSO phase is found to be a mixture of 18R and 14H structures, which is consistent with the results reported in the literature [17-20]. Another intermetallic phase in the $\text{Mg}_{85.8}\text{Y}_{7.1}\text{Zn}_{7.1}$ alloy is identified as the W-phase with an fcc structure ($a = \sim 0.68$ nm), as shown in Fig. 3. The EDX analysis suggests that the W-phase has a near-stoichiometric composition close to MgZn_2Y .

The LPSO phase in the $\text{Mg}_{85.8}\text{Y}_{7.1}\text{Ni}_{7.1}$ alloy is found to solely consist of the 18R structure, as shown in Fig. 4. Similar results have been reported for $\text{Mg}_{94}\text{Y}_4\text{Ni}_2$ [11] and $\text{Mg}_{90.5}\text{Y}_{3.25}\text{Ni}_{6.25}$ alloys [21]. No stoichiometric composition, however, was provided for the LPSO phase in those reports. The current EDX analysis suggests that the Mg–Y–Ni LPSO phase may have a stoichiometric composition of $\text{Mg}_{12}\text{Y}_2\text{Ni}$. Beside the LPSO phase, two other intermetallic phases are present in the $\text{Mg}_{85.8}\text{Y}_{7.1}\text{Ni}_{7.1}$ alloy; one has smaller particles but is more abundant and the other has larger particles, but is less frequently observed. Fig. 5 shows the identification of the intermetallic phase with the smaller particles. This intermetallic phase contains mainly Mg and Ni, with a small amount of Y. The electron diffraction analysis suggests that this phase should be Mg_2Ni with a hexagonal structure ($a = 0.52$ nm, $c = 1.32$ nm), which has also been identified in a $\text{Mg}_{83}\text{Ni}_6\text{Zn}_5\text{Y}_6$ alloy [13]. The less frequently observed intermetallic phase contains a significant amount of Mg, Ni and Y, and the electron diffraction pattern can be indexed as MgNi_4Y , as shown in Fig. 6. The MgNi_4Y phase has been reported to have a C15b-type Laves phase structure (fcc, $a = 0.701$ nm) [22].

3.2. Microstructure after extrusion or ECAP

The typical SEM images of the $\text{Mg}_{85.8}\text{Y}_{7.1}\text{Zn}_{7.1}$ and $\text{Mg}_{85.8}\text{Y}_{7.1}\text{Ni}_{7.1}$ alloys after the extrusion or ECAP at 450°C are shown in Fig. 7. As compared with the homogenised state (Fig. 1), the following changes in microstructure due to the large deformation by extrusion or ECAP are noted: (i) the LPSO lamellae are no longer straight and become bent or wavy; (ii) features suggesting the occurrence of kinking are common in the LPSO lamellae; (iii) some intermetallic particles appear to be refined due to fragmentation, however there appear to be no changes in phase constituents after the deformation; the $\text{Mg}_{85.8}\text{Y}_{7.1}\text{Zn}_{7.1}$ alloy still consists of the LPSO phase, the W-phase, and α -Mg whilst the phases in $\text{Mg}_{85.8}\text{Y}_{7.1}\text{Ni}_{7.1}$ alloys include the LPSO phase, Mg_2Ni , MgNi_4Y , and α -Mg. It is worth mentioning that, despite the large strain induced by the deformation, especially in the case of ECAP, no obvious dynamic recrystallisation (DRX) of the LPSO phase was revealed by the initial SEM observations in either alloy after processing. This appears to be consistent with a previous report [23] that conventional DRX process is not expected to be operative for the LPSO phase due to its high thermal stability [24]. Further investigation by TEM examinations, however, provided evidence for the occurrence of DRX in the LPSO phase in both alloys.

Fig. 8 shows the typical TEM micrographs of the extruded or ECAP processed $\text{Mg}_{85.8}\text{Y}_{7.1}\text{Zn}_{7.1}$ and $\text{Mg}_{85.8}\text{Y}_{7.1}\text{Ni}_{7.1}$ alloys, with special attention paid to the LPSO phase. Although the microstructures are comprised predominantly by large, slightly bent LPSO lamellae, it is interesting that fine LPSO grains (around $0.3\text{ }\mu\text{m}$) are also observed. These fine LPSO grains are considered to form via DRX during the hot extrusion or ECAP. Similar DRX grains have been observed in extruded $\text{Mg}_{89}\text{Zn}_4\text{Y}_7$ alloy [25]. It was suggested [25] that the DRX of the LPSO phase takes place only locally, in the vicinity of grain boundaries with strong stress concentrations. However, there is evidence in the current study that the DRX likely occurs preferentially in the kinked or heavily bent regions of the LPSO phase, as shown in Fig. 9. A high density of dislocations are accumulated in these deformed regions, especially at the kink boundaries; the rearrangement of these dislocations leads to dynamic recovery and DRX during the extrusion or ECAP. This process is likely very slow and, consequently, only a small fraction of the LPSO phase undergoes DRX.

3.3. Compression tests

Mechanical properties of the $\text{Mg}_{85.8}\text{Y}_{7.1}\text{Zn}_{7.1}$ and $\text{Mg}_{85.8}\text{Y}_{7.1}\text{Ni}_{7.1}$ alloys were evaluated for the extruded and ECAP processed rods, rather than the cast and homogenised ingots because, even in compression, the alloys in the cast and homogenised state exhibited very poor strength ($\sim 100\text{ MPa}$) and extremely low ductility ($\sim 1\%$). The extruded and ECAP processed rods also exhibited extremely low ductility when tested in tension. As a consequence, compression tests were conducted on the extruded and ECAP processed rods for evaluation of mechanical properties.

The strength of the alloys under compression was significantly improved after the extrusion or ECAP. Fig. 10 shows the room temperature compression stress-strain curves of the extruded and ECAP processed rods and the measured yield strength, compressive strength and ductility (measured as the strain at fracture) are presented in Table 1. For both processing routes, the substitution of Zn with Ni is seen to significantly increase the strength, but this increment in strength is accompanied with a loss in ductility. It is worth noting that the strength after the extrusion is much higher than that after the ECAP for both $\text{Mg}_{85.8}\text{Y}_{7.1}\text{Zn}_{7.1}$ and $\text{Mg}_{85.8}\text{Y}_{7.1}\text{Ni}_{7.1}$ alloy and there appears to be no significant difference in ductility between the two processing routes.

4. Discussion

4.1. Phases in the $Mg_{85.8}Y_{7.1}Zn_{7.1}$ and $Mg_{85.8}Y_{7.1}Ni_{7.1}$ alloys

Consistent with the reported in the literature [17-20], a mixture of 18R and 14H structures is found for the LPSO phase in the cast $Mg_{85.8}Y_{7.1}Zn_{7.1}$ alloy after homogenisation at 450°C for 24 h. This is because 18R is not thermally stable and undergoes a transition to 14H upon heat treatment. Since the LPSO phase in the $Mg_{85.8}Y_{7.1}Ni_{7.1}$ alloys consists solely of the 18R structure, it can be deduced that the substitution of Zn with Ni probably has a stabilising effect on the 18R variant of the LPSO phase.

The stoichiometry of the W-phase is commonly accepted to be $Mg_3Zn_3Y_2$ [15, 26-28], but a recent study [29] suggested that the W-phase should have an $L2_1$ -type crystal structure with a stoichiometric composition of $MgZn_2Y$, with Mg and Zn being substitutable for each other. The stoichiometric composition $MgZn_2Y$ has also been adopted for the W-phase in a recent thermodynamic calculation of the Mg–Y–Zn phase diagram [30]. The EDX analysis in the present study is in support of the stoichiometric composition $MgZn_2Y$ for the W-phase, but there is no evidence that this phase could have the $L2_1$ -type crystal structure.

For the cast $Mg_{85.8}Y_{7.1}Ni_{7.1}$ alloy after homogenisation at 450 °C for 24 h, the phases identified include 18R LPSO, Mg_2Ni , $MgNi_4Y$, and α -Mg. A study [31] on phase equilibria of Mg–Ni–Y system reported 18R LPSO, Mg_2Ni , and $MgNi_4Y$ for a similar alloy after heat treatment at 400°C or 500°C for 720 h. However, by applying a heat treatment at 400°C for 1500 h or 500°C for 768 h, another study [32] suggested that 14H should be the equilibrium LPSO phase in the Mg–Ni–Y system in a wide temperature range. The equilibrium 14H phase was reported [32] to have a stoichiometric composition close to $Mg_{91}Ni_4Y_5$. Thus, it cannot be ruled out that the 18R LPSO observed in the current $Mg_{85.8}Y_{7.1}Ni_{7.1}$ alloy is not the equilibrium phase and will eventually transform to 14H LPSO. However, given that the 18R LPSO phase in the Mg–Y–Ni system has enhanced thermal stability, it can be conjectured that in that case the transformation to 14H LPSO would be sluggish.

4.2. Mechanical properties of the $Mg_{85.8}Y_{7.1}Zn_{7.1}$ and $Mg_{85.8}Y_{7.1}Ni_{7.1}$ alloys

The substitution of Zn with Ni is found to significantly improve the strength of the $\text{Mg}_{85.8}\text{Y}_{7.1}\text{Zn}_{7.1}$ alloy, but have a negative effect on the ductility. Since the $\text{Mg}_{85.8}\text{Y}_{7.1}\text{Ni}_{7.1}$ alloy has a distinctly higher fraction of the LPSO phase and thus a lower fraction of the α -Mg phase than the $\text{Mg}_{85.8}\text{Y}_{7.1}\text{Ni}_{7.1}$ alloy, the observed differences in the mechanical properties are considered to be mainly due to the difference in the phase composition of the alloys.

Given that the equivalent strain introduced by the four-pass ECAP is almost twice as large as that introduced by the extrusion, it is surprising to see that the strengths of the ECAP processed alloys are actually much lower than those of the extruded ones. However, there appear no apparent differences in microstructure between the alloys processed by the two processing routes, as revealed by the microstructural observations (Figs. 7-9). The differences in strength between the ECAP processed alloys and the extruded alloys are thus considered to be related to the different textures developed by the two processing routes. Mg alloys normally have a strong fibre texture after extrusion [33] while a more random texture is usually developed after ECAP, especially by Route B_C [34]. Moreover, the basal planes tend to be oriented at $\sim 45^\circ$ to the pressing direction for Mg alloys after Route B_C ECAP [35], which makes basal slip easy to activate if a compressive load is applied along the pressing direction, as is the case for this study. Further analysis of the recrystallization and grain structure by electron backscatter diffraction (EBSD) will help get a better understanding of the observed differences in strength between the ECAP processed alloys and the extruded counterparts.

5. Conclusions

The comparative study of the microstructure and mechanical properties of $\text{Mg}_{85.8}\text{Y}_{7.1}\text{Zn}_{7.1}$ and $\text{Mg}_{85.8}\text{Y}_{7.1}\text{Ni}_{7.1}$ alloys conducted in this work leads to the following conclusions:

- (1) Both alloys are composed primarily of the LPSO phase, with small amounts of α -Mg and other intermetallic phases, i.e. MgZn_2Y (W-phase) for the $\text{Mg}_{85.8}\text{Y}_{7.1}\text{Zn}_{7.1}$ alloy and Mg_2Ni as well as MgNi_4Y for the $\text{Mg}_{85.8}\text{Y}_{7.1}\text{Ni}_{7.1}$ alloy. The $\text{Mg}_{85.8}\text{Y}_{7.1}\text{Ni}_{7.1}$ alloy contains a markedly higher fraction of the LPSO phase and a lower fraction of the α -Mg phase than its counterpart.

- (2) The LPSO phase in the homogenised $\text{Mg}_{85.8}\text{Y}_{7.1}\text{Zn}_{7.1}$ alloy has mixed 18R and 14H structures while it consists of 18R only in the homogenised $\text{Mg}_{85.8}\text{Y}_{7.1}\text{Ni}_{7.1}$ alloy.
- (3) The substitution of Zn with Ni significantly improves the strength of the $\text{Mg}_{85.8}\text{Y}_{7.1}\text{Zn}_{7.1}$ alloy, but this improvement is accompanied by a drop in ductility. The differences in mechanical properties are attributed to the difference in the volume fraction of the LPSO phase and, accordingly, the α -Mg phase in the alloys.

Acknowledgements

This work was supported by the U.S. Army International Technology Center – Pacific and the US Army Research Laboratory under Grant FA5209-11-T-0043, with Dr. Christopher Drew and Dr. Vincent Hammond as program manager, respectively. S.N. Mathaudhu gratefully acknowledges the support of the US National Science Foundation under Grant 1554632. Monash Centre for Electron Microscopy (MCEM) is acknowledged for access to experimental facilities. The authors would also like to thank Sam Gao for conducting the compression tests.

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Table 1 Room temperature compression test results for extruded and ECAP processed $\text{Mg}_{85.8}\text{Y}_{7.1}\text{Zn}_{7.1}$ and $\text{Mg}_{85.8}\text{Y}_{7.1}\text{Ni}_{7.1}$ alloys

Alloy	Extruded			ECAP Processed		
	Yield Strength (MPa)	Compressive Strength (MPa)	Ductility (%)	Yield Strength (MPa)	Compressive Strength (MPa)	Ductility (%)
$\text{Mg}_{85.8}\text{Y}_{7.1}\text{Zn}_{7.1}$	467 ± 15	650 ± 5	12.3 ± 0.6	365 ± 26	546 ± 48	10.4 ± 4.2
$\text{Mg}_{85.8}\text{Y}_{7.1}\text{Ni}_{7.1}$	602 ± 3	763 ± 25	7.8 ± 1.5	453 ± 34	670 ± 46	8.6 ± 1.1

Fig. 1. SEM backscattered electron images showing the microstructures of (a) $\text{Mg}_{85.8}\text{Y}_{7.1}\text{Zn}_{7.1}$ and (b) $\text{Mg}_{85.8}\text{Y}_{7.1}\text{Ni}_{7.1}$ alloys homogenised at 450°C for 24 h. The phases present in the microstructure are labelled. The identification of these phases is based on TEM analysis.

Fig. 2. TEM bright field image and selected area electron diffraction patterns showing the LPSO phase with a mixture of 18R and 14H structures in the homogenised $\text{Mg}_{85.8}\text{Y}_{7.1}\text{Zn}_{7.1}$ alloy. The weak streaks arising from ordering are indicated by arrows in the diffraction patterns.

Fig. 3. Identification of W-phase in the homogenised $\text{Mg}_{85.8}\text{Y}_{7.1}\text{Zn}_{7.1}$ alloy: (a) TEM bright field image and selected area electron diffraction pattern and (b) EDX analysis. The W-phase is identified to have an fcc structure with a lattice parameter of $\sim 0.68\ \text{nm}$.

Fig. 4. Characterization of the LPSO phase in the homogenised $\text{Mg}_{85.8}\text{Y}_{7.1}\text{Ni}_{7.1}$ alloy: (a) TEM bright field image, (b) HRTEM image, (c) selected area electron diffraction pattern and (d) EDX analysis. The LPSO phase consists of the 18R structure only. The weak streaks arising from ordering are indicated by arrows in the diffraction patterns.

Fig. 5. Identification of the Mg₂Ni phase in the homogenised Mg_{85.8}Y_{7.1}Ni_{7.1} alloy: (a) TEM bright field image and selected area electron diffraction pattern and (b) EDX analysis. A small amount of Y is detected in the Mg₂Ni phase.

Fig. 6. Identification of the MgNi_4Y phase in the homogenised $\text{Mg}_{85.8}\text{Y}_{7.1}\text{Ni}_{7.1}$ alloy: (a) TEM bright field image and selected area electron diffraction pattern and (b) EDX analysis.

Fig. 7. SEM backscattered electron images showing the microstructures of (a) extruded $\text{Mg}_{85.8}\text{Y}_{7.1}\text{Zn}_{7.1}$, (b) extruded $\text{Mg}_{85.8}\text{Y}_{7.1}\text{Ni}_{7.1}$, (c) ECAP processed $\text{Mg}_{85.8}\text{Y}_{7.1}\text{Zn}_{7.1}$ and (d) ECAP processed $\text{Mg}_{85.8}\text{Y}_{7.1}\text{Ni}_{7.1}$ alloys. The arrows indicate kink-like features. There are no discernible changes in the phase constituents as a result of deformation.

Fig. 8. TEM bright field images showing the microstructures of (a) extruded $\text{Mg}_{85.8}\text{Y}_{7.1}\text{Zn}_{7.1}$, (b) extruded $\text{Mg}_{85.8}\text{Y}_{7.1}\text{Ni}_{7.1}$, (c) ECAP processed $\text{Mg}_{85.8}\text{Y}_{7.1}\text{Zn}_{7.1}$ and (d) ECAP processed $\text{Mg}_{85.8}\text{Y}_{7.1}\text{Ni}_{7.1}$ alloys. Note the occurrence of fine grains of the LPSO phase in the microstructures as a result of DRX.

Fig. 9. TEM bright field images of the kinked or heavily bent LPSO regions in (a) extruded $\text{Mg}_{85.8}\text{Y}_{7.1}\text{Ni}_{7.1}$ and (b) ECAP processed $\text{Mg}_{85.8}\text{Y}_{7.1}\text{Zn}_{7.1}$ alloys. Note the development of fine grains in these regions due to DRX.

Fig. 10. Room temperature compression test curves for (a) extruded and (b) ECAP processed $\text{Mg}_{85.8}\text{Y}_{7.1}\text{Zn}_{7.1}$ and $\text{Mg}_{85.8}\text{Y}_{7.1}\text{Ni}_{7.1}$ alloys.