

Full Length Article

Extruded $\text{Mg}_{97.37}\text{Zn}_{1.88}\text{Y}_{0.75}$ alloy with both high thermal conductivity and mechanical strength

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Received 17 July 2025; received in revised form 9 September 2025; accepted 23 September 2025

Available online 22 October 2025

Abstract

In our previous study, we observed that the $\text{Mg}_{97.37}\text{Zn}_{1.88}\text{Y}_{0.75}$ (at%) alloy, composed of α -Mg, the W phase ($\text{Mg}_3\text{Zn}_3\text{Y}_2$), and a small quantity of the I phase ($\text{Mg}_3\text{Zn}_6\text{Y}$) exhibited an extremely high thermal conductivity of $141 \text{ Wm}^{-1}\text{K}^{-1}$ attributable to the precipitation of a large amount of the fine W phase through heat treatment at 633 K for 15 h. In the present study, we examined how extrusion conditions affect the thermal conductivity and mechanical properties of the $\text{Mg}_{97.37}\text{Zn}_{1.88}\text{Y}_{0.75}$ alloy with high thermal conductivity. As a result, an extruded Mg alloy was developed, exhibiting excellent properties, including a thermal conductivity of $131 \text{ Wm}^{-1}\text{K}^{-1}$, a tensile strength of 361 MPa, and an elongation of 9.7%. These superior mechanical properties were obtained by optimizing extrusion conditions, thereby inducing dynamic recrystallization during the process to refine the α -Mg matrix and causing the dynamically recrystallized area, which consisted of grains with lower strain and random orientation, to expand. Furthermore, a decline in thermal conductivity by 9–11 $\text{Wm}^{-1}\text{K}^{-1}$ was observed following extrusion, independent of the specific processing parameters.

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Keywords: Mg–Zn–Y alloy; Thermal conductivity; Mechanical properties; Extrusion processing.

1. Introduction

Since Mg alloys are the lightest metal material in practical use, they have garnered significant interest for their potential in reducing the weight of transportation vehicles. Such weight reduction contributes significantly to energy conservation and emission reduction efforts [1,2]. In recent years, the rapid development of aerospace systems, high-speed trains, autonomous driving systems, and others has resulted in an increase in both the quantity and layout density of high-power electromagnetic components, leading to greater heat generation during equipment operation. If the large amount of heat generated cannot be dissipated promptly, it

will lead to excessively high equipment temperature, which will severely affect stability and safety when operating the equipment [3]. However, compared with Al alloys, commercial Mg alloys have lower mechanical strength and thermal conductivity (TC), limiting their application as structural components in vehicle-mounted high-temperature equipment and on-board electronics.

To solve these problems, there have been many attempts to develop Mg alloys with the combination of high TC, strength, and ductility. In general, various strengthening methods, including solid solution strengthening [4,5], grain refinement [6–9], work hardening [10–12], precipitation strengthening [6,13,14], dispersion hardening [12], and most recently, kink banding strengthening [15–19], can be used to develop high-strength Mg alloys. However, a trade-off typically exists between yield strength (YS) and elongation (EI). In addition, these strengthening methods also introduce lattice defects such as solute atoms, second phases, grain boundaries, twin boundaries, stacking faults, and dislocations. These de-

Peer review under the responsibility of Chongqing University.

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fects enhance the probability of scattering of free electrons and phonons, which serve as carriers for heat conduction. As a result, the mean free path of the heat carriers decreases, which in turn lowers TC. Consequently, a similar trade-off also exists between YS and TC [9,20–24]. Therefore, studies on high-TC Mg alloys primarily concentrate on the selection of alloying elements and are carried out on the basis of the principle of using alloys with fewer elements added and applying manufacturing methods such as die-casting [21–29]. Recently, Mg alloys that exhibit both high TC and relatively high strength have been successfully produced by extruding alloys with a limited addition of alloying elements. Thus far, Yuan and coworkers have developed Mg–Zn–Mn ($\text{Zn} \leq 3.15$ at%, $\text{Mn} \leq 0.47$ at%) alloys with a TC of $120\text{--}125 \text{ Wm}^{-1}\text{K}^{-1}$, a YS of 211–288 MPa, and an El of 14.8%–16.0% [27,28]. Wang et al. developed the Mg–1.14Zn–0.16Zr (at%) alloy with a TC of $126 \text{ Wm}^{-1}\text{K}^{-1}$, a YS of 255 MPa, and an El of 7% [29]. Zhong et al. also developed the Mg–0.02Mn–0.05Ce (at%) alloy with a TC of $126.9 \text{ Wm}^{-1}\text{K}^{-1}$, a YS of 295.9 MPa, and an El of 9.6% [30]. Although the restricted addition of alloying elements contributes to enhancing thermal conductivity, the resulting alloys suffer from insufficient yield strength, typically below 300 MPa.

In our previous research [31], we optimized the composition and heat treatment conditions of a Mg–Zn–Y alloy. This alloy contains a Zn–Y pair characterized by a highly negative mixing enthalpy of -31 kJ/mol . Through this optimization, the W phase ($\text{Mg}_3\text{Zn}_3\text{Y}_2$), which consumes alloying elements, particularly Y, a key element affecting the thermal conductivity of Mg alloys, was more extensively crystallized and precipitated. Consequently, the α -Mg matrix was significantly purified. As a result, we have successfully developed a casting heat-treated $\text{Mg}_{97.37}\text{Zn}_{1.88}\text{Y}_{0.75}$ alloy that exhibits an extremely high thermal conductivity reaching $141 \text{ Wm}^{-1}\text{K}^{-1}$. The alloy is composed of the α -Mg phase, the W phase, and a small amount of the I phase ($\text{Mg}_3\text{Zn}_6\text{Y}$). Jiang and coworkers [32,33] applied high-temperature extrusion processing to a Mg–Zn–Y–Zr–Ca alloy, in which the primary constituent phase was the W phase. As a result of this processing, the nanoscale fine W phase precipitated within the matrix. Simultaneously, the initially crystallized coarse W phase was fractured and uniformly dispersed throughout the matrix. This microstructural evolution facilitated the development of a bimodal grain structure, consisting of fine dynamically recrystallized grains and coarse non-recrystallized grains, resulting in the high strength and moderate ductility of the alloy. Therefore, by properly extruding the $\text{Mg}_{97.37}\text{Zn}_{1.88}\text{Y}_{0.75}$ alloy, it is also possible to obtain high strength and ductility while maintaining high thermal conductivity.

Therefore, we took the high-thermal-conductivity, casting heat-treated $\text{Mg}_{97.37}\text{Zn}_{1.88}\text{Y}_{0.75}$ alloy as the subject in the present study, with the aim of obtaining high mechanical strength while maintaining the high thermal conductivity. The effects of extrusion conditions on the microstructure and properties of the extruded $\text{Mg}_{97.37}\text{Zn}_{1.88}\text{Y}_{0.75}$ alloy were systematically studied.

2. Experimental procedure

$\text{Mg}_{97.37}\text{Zn}_{1.88}\text{Y}_{0.75}$ (at%) alloy samples were fabricated using pure Mg, Zn, and Y with a purity of 99.9 mass% using a high-frequency induction dissolution furnace. The alloy melt was retained in a carbon crucible at 1023 K for 15 min, followed by a 15 min stirring process using a carbon rod, after which the molten alloy was transferred to a mild steel mold for solidification. The entire process was carried out in Ar atmosphere. The as-cast $\text{Mg}_{97.37}\text{Zn}_{1.88}\text{Y}_{0.75}$ alloys were subjected to heat treatment at 633 K for 15 h using a tubular furnace. After heat treatment, the samples were air-cooled at a rate of $1.2 \text{ K}\cdot\text{s}^{-1}$. For extrusion processing, the heat-treated ingots were processed into round bars with the dimensions of $\phi 29 \text{ mm} \times \text{L}60 \text{ mm}$. Subsequently, extrusion processing was carried out using an 80t hydraulic extrusion press under various conditions, including extrusion ratios from 5 to 20, ram speeds from 0.5 to 5.5 $\text{mm}\cdot\text{s}^{-1}$, and temperatures from 523 to 673 K.

The thermal conductivity of the extruded $\text{Mg}_{97.37}\text{Zn}_{1.88}\text{Y}_{0.75}$ alloys was quantitatively determined at 298 K by a laser flash analysis technique using the LFA 467 HyperFlash apparatus. The thermal conductivity of all the extruded alloys was measured in both the extrusion direction (ED) and the transverse direction (TD). The samples were processed into discs measuring $\phi 9 \text{ mm} \times \text{L}2 \text{ mm}$ and ground using #2000 SiC paper until both surfaces were flat, smooth, and free of visible defects. Pyroceram samples were used as reference samples to calibrate the thermal diffusivity and specific heat capacity of the instrument. Before the measurements, the samples including the reference were sprayed with Dry Graphite Film Spray to enhance the absorption of flash light pulses. During measurements, a Xenon flash lamp was employed to generate the light pulses, and the surface temperature was recorded with an InSb sensor. The mechanical properties of the extruded alloys were evaluated at room temperature. An Instron 5584 testing machine was employed for the measurements. The elongation of samples was evaluated using an extensometer (AVE 2) under a strain rate of $5 \times 10^{-4} \text{ s}^{-1}$. Dog-bone-type samples were used in tensile testing. During the tensile testing, the loading axis was oriented parallel to the extrusion direction (ED). To reduce experimental errors, thermal conductivity and stress–strain curves were obtained from at least three samples.

The phase compositions of the extruded $\text{Mg}_{97.37}\text{Zn}_{1.88}\text{Y}_{0.75}$ alloys were ascertained using an XRD device (Bruker, D8 Discover). XRD patterns were collected over a 2θ range from 20° to 90° , with a scanning increment of 0.02° per second. The microstructure of the alloys was observed by SEM (JIB-4601F). Crystallographic orientation analysis was performed by EBSD using OIM. To assess grain size, internal strain distribution, and degree of orientation, inverse polar figure (IPF) mapping, kernel average misorientation (KAM) mapping, and polar figure imaging were employed. Each scanned region covered an area of $100 \times 200 \mu\text{m}^2$ with a step size of $0.2 \mu\text{m}$ to ensure high spatial resolution.

3. Results

3.1. Effects of extrusion conditions on properties of $Mg_{97.37}Zn_{1.88}Y_{0.75}$ alloy

Figs. 1(a), (b), and (c) respectively illustrate the effects of extrusion ratio (R), extrusion ram speed (V_r), and extrusion temperature (T_e) on the thermal conductivity of the extruded $Mg_{97.37}Zn_{1.88}Y_{0.75}$ alloys. The results show that the thermal conductivity of the extruded $Mg_{97.37}Zn_{1.88}Y_{0.75}$ alloys is almost independent of the extrusion conditions. After extrusion processing, it decreased by about $9\text{--}11\text{ Wm}^{-1}\text{K}^{-1}$ regardless of the extrusion conditions, but still remained very high at a value above $130\text{ Wm}^{-1}\text{K}^{-1}$.

Figs. 2(a), (b), and (c) respectively illustrate the effects of extrusion ratio, extrusion ram speed, and extrusion temperature on the mechanical properties of the extruded $Mg_{97.37}Zn_{1.88}Y_{0.75}$ alloys. As shown in Fig. 2(a), at a V_r of $1.0\text{ mm}\cdot\text{s}^{-1}$ and a T_e of 523 K , as the extrusion ratio was raised from $R5$ to $R20$, the tensile yield strength increased initially and reached the maximum at the extrusion ratio of 10, and then gradually decreased. In contrast, the elongation increased consistently with each increment in extrusion ratio. When the extrusion ratio was above 10, the alloy simultaneously exhibited a very high tensile yield strength ($\sigma_{0.2} > 350\text{ MPa}$) and a sufficient ductility ($\delta > 5\%$). As shown in Fig. 2(b), at an R

of 15 and a T_e of 523 K , as the extrusion ram speed was raised from 0.5 to $5.5\text{ mm}\cdot\text{s}^{-1}$, the tensile yield strength decreased significantly, whereas the elongation increased markedly. At an extrusion ram speed of $1.0\text{--}2.5\text{ mm}\cdot\text{s}^{-1}$, the alloy simultaneously exhibited a high yield strength ($\sigma_{0.2} > 300\text{ MPa}$) and a high ductility ($\delta > 9\%$). As shown in Fig. 2(c), at an R of 15 and a V_r of $1.0\text{ mm}\cdot\text{s}^{-1}$, as the extrusion temperature was raised from 523 to 673 K , the tensile yield strength decreased monotonically and the elongation was almost unchanged. When the temperature was below 623 K , high yield strength ($\sigma_{0.2} > 300\text{ MPa}$) and high elongation ($\delta > 9\%$) were obtained simultaneously.

Fig. 3(a) illustrates the relationship between tensile yield strength and elongation for $Mg_{97.37}Zn_{1.88}Y_{0.75}$ alloys. The optimal extrusion conditions for obtaining high tensile yield strength while having an elongation of about 10% are $R = 15$, $V_r = 1.0\text{ mm}\cdot\text{s}^{-1}$, and $T_e = 523\text{ K}$. Fig. 3(b) illustrates the relationship between thermal conductivity and tensile yield strength for the extruded $Mg_{97.37}Zn_{1.88}Y_{0.75}$ alloys. In this study, the thermal conductivity of the extruded $Mg_{97.37}Zn_{1.88}Y_{0.75}$ alloys was found to be independent of extrusion conditions. Extrusion processing resulted in a decrease of approximately $10\text{ Wm}^{-1}\text{K}^{-1}$ in thermal conductivity for all alloys under all the investigated conditions. Consequently, the alloy extruded under the conditions yielding the most balanced mechanical properties still retained an ex-

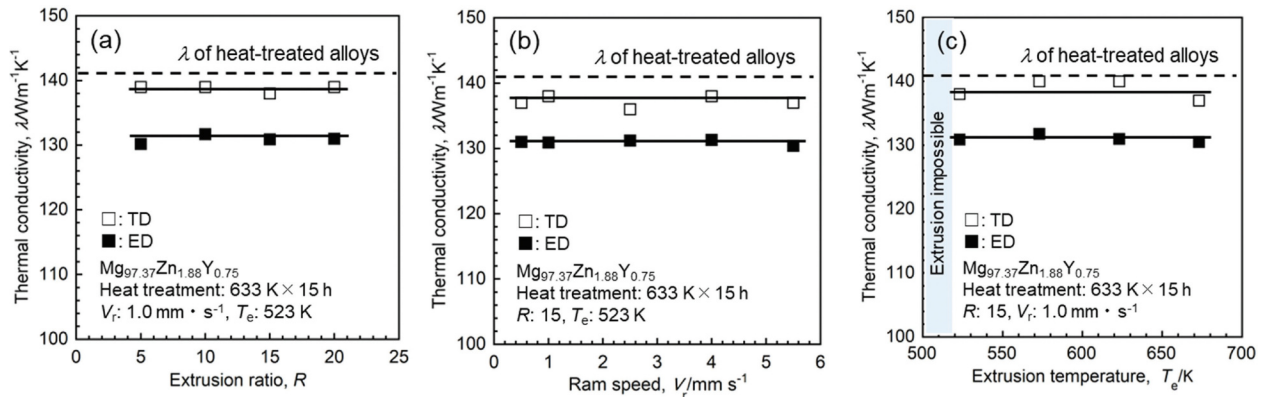


Fig. 1. Effects of (a) extrusion ratio (R), (b) extrusion ram speed (V_r), and (c) extrusion temperature (T_e) on the thermal conductivity of extruded $Mg_{97.37}Zn_{1.88}Y_{0.75}$ (at%) alloys.

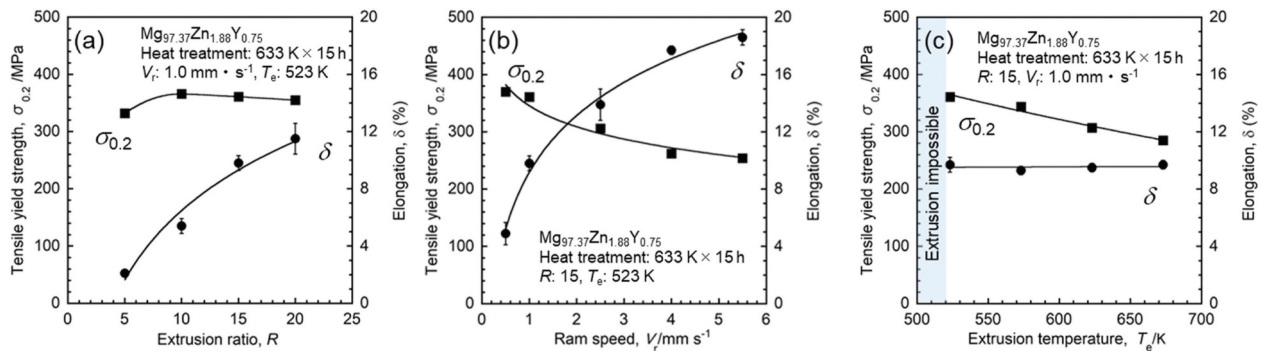


Fig. 2. Effects of (a) extrusion ratio (R), (b) extrusion ram speed (V_r), and (c) extrusion temperature (T_e) on the tensile yield strength and elongation of extruded $Mg_{97.37}Zn_{1.88}Y_{0.75}$ (at%) alloys.

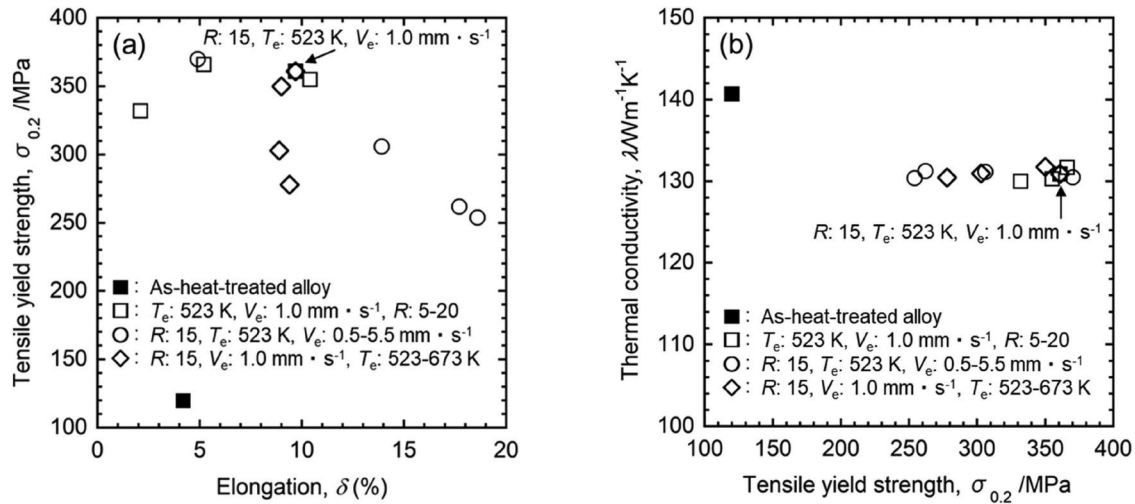


Fig. 3. Relationships between (a) tensile yield strength and elongation, and (b) thermal conductivity and tensile yield strength of extruded Mg_{97.37}Zn_{1.88}Y_{0.75} (at%) alloys.

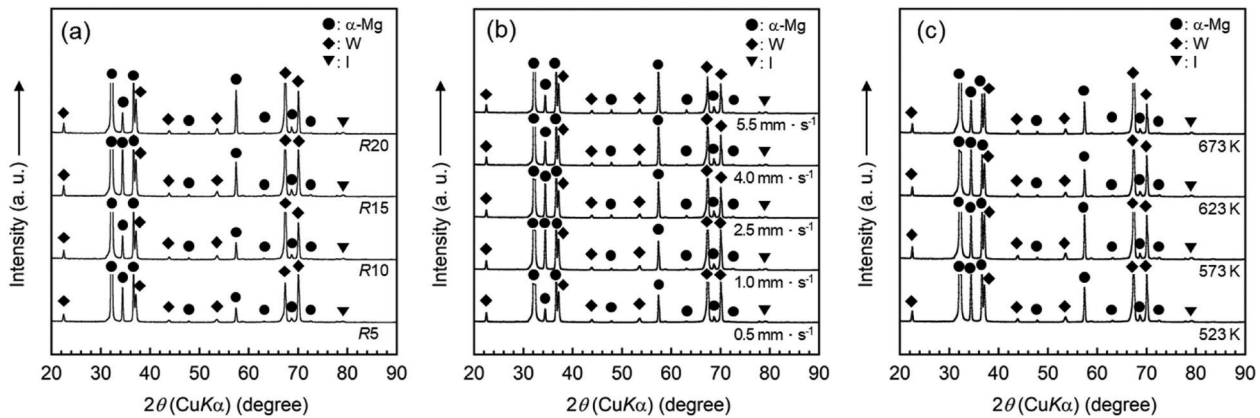


Fig. 4. Effects of (a) extrusion ratio (R), (b) extrusion ram speed (V_r), and (c) extrusion temperature (T_e) on X-ray diffraction patterns of extruded Mg_{97.37}Zn_{1.88}Y_{0.75} (at%) alloys.

cellent thermal conductivity of 131 Wm⁻¹K⁻¹. These results show that the extruded Mg_{97.37}Zn_{1.88}Y_{0.75} alloy prepared under the optimal extrusion conditions simultaneously achieved a thermal conductivity as high as 131 Wm⁻¹K⁻¹, a high yield strength reaching 361 MPa, and a notable elongation of 9.7%.

3.2. Microstructure evolution of extruded Mg_{97.37}Zn_{1.88}Y_{0.75} alloys

Figs. 4(a), (b), and (c) show the XRD patterns obtained from the extruded Mg_{97.37}Zn_{1.88}Y_{0.75} alloys under different conditions. Irrespective of extrusion conditions, the constituent phases of the extruded Mg_{97.37}Zn_{1.88}Y_{0.75} alloys did not change. Similarly to the heat-treated alloys, the XRD patterns of the extruded alloys exhibited diffraction peaks corresponding to the α -Mg and W phases, along with a weak peak associated with the I phase.

Fig. 5 shows the SEM images obtained from the extruded Mg_{97.37}Zn_{1.88}Y_{0.75} alloys under different conditions. Fig. 6 illustrates the effects of the extrusion ratio, extrusion ram

speed, and extrusion temperature on the dispersion of compound bands and the particle size of compounds, ascertained from SEM images, in extruded Mg_{97.37}Zn_{1.88}Y_{0.75} alloys. As the extrusion ratio was raised from R5 to R20, the dispersion of compound bands increased from 0.06 to 0.1 μm^{-1} , whereas the particle size of compounds remained almost unchanged at approximately 0.40 μm . Moreover, the dispersion of compound bands and the particle size of compounds were almost unaffected by extrusion ram speed and extrusion temperature, and remained at around 0.09 μm^{-1} and 0.40 μm , respectively.

Fig. 7 shows the IPF maps of the extruded Mg_{97.37}Zn_{1.88}Y_{0.75} alloys under different extrusion conditions. Fig. 8 illustrates the effects of extrusion conditions on the area fraction and grain size of dynamic recrystallized (DRXed) α -Mg, ascertained from the IPF maps, in extruded Mg_{97.37}Zn_{1.88}Y_{0.75} alloys. When the extrusion ratio increased from R5 to R20, the grain size of the DRXed α -Mg phase remained nearly constant at approximately 1.5 μm . However, the area fraction of the DRXed α -Mg significantly increased from 23 to 75%. With the extrusion ram speed increasing

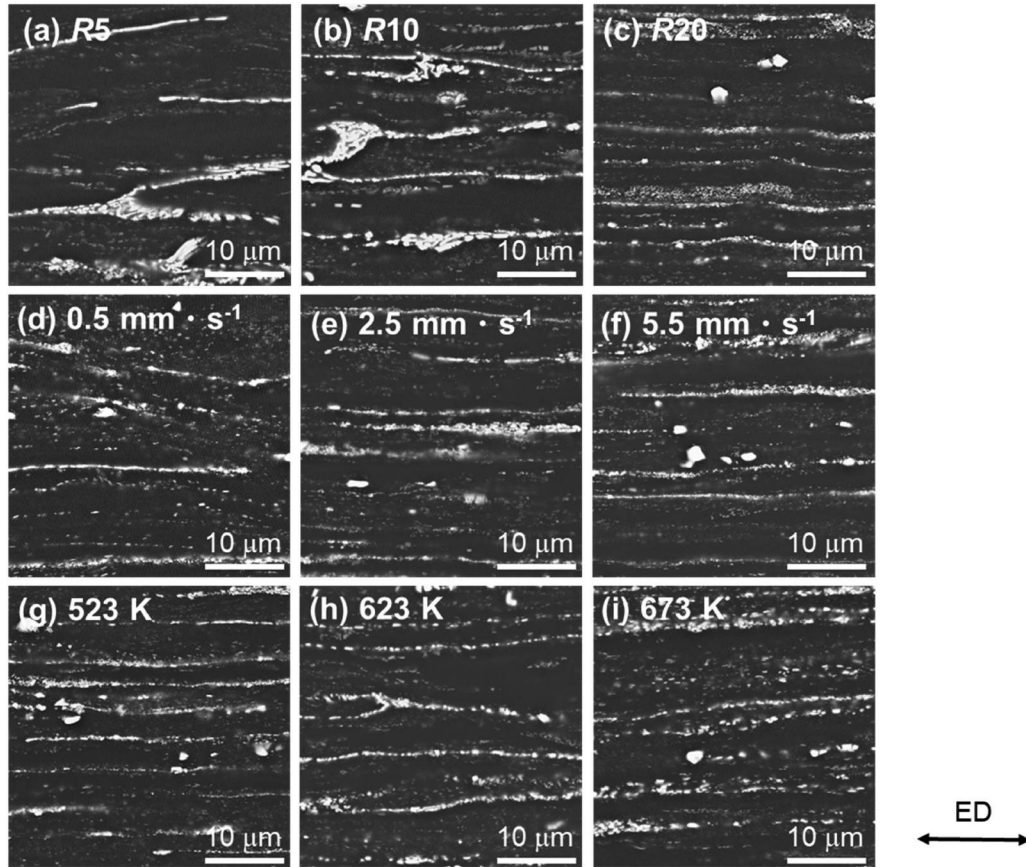


Fig. 5. SEM images showing effects of extrusion ratio [(a) $R5$, (b) $R10$, (c) $R20$], extrusion ram speed [(d) $0.5 \text{ mm} \cdot \text{s}^{-1}$, (e) $2.5 \text{ mm} \cdot \text{s}^{-1}$, (f) $5.5 \text{ mm} \cdot \text{s}^{-1}$], and extrusion temperature [(g) 523 K , (h) 623 K , (i) 673 K] on extruded $\text{Mg}_{97.37}\text{Zn}_{1.88}\text{Y}_{0.75}$ (at %) alloys.

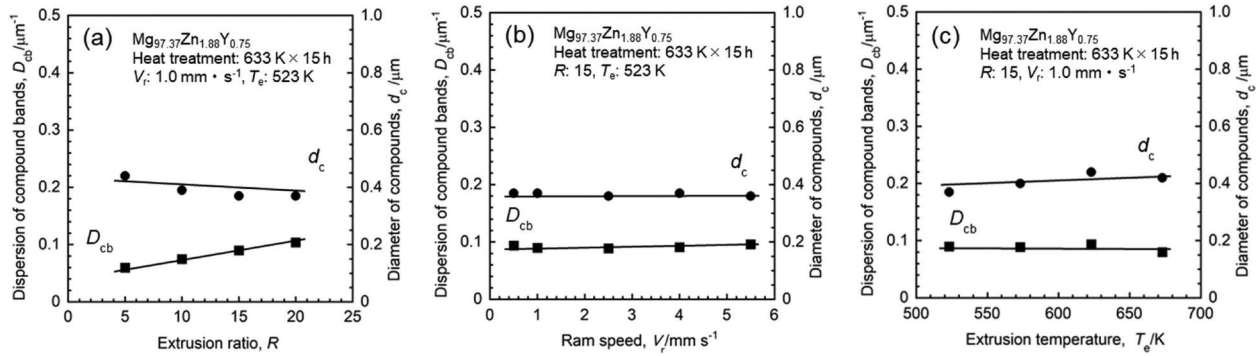


Fig. 6. Effects of (a) extrusion ratio (R), (b) extrusion ram speed (V_r), and (c) extrusion temperature (T_e) on the dispersion of compound bands and particle size of compounds in extruded $\text{Mg}_{97.37}\text{Zn}_{1.88}\text{Y}_{0.75}$ (at %) alloys.

from 0.5 to $5.5 \text{ mm} \cdot \text{s}^{-1}$, both the grain size and area fraction of the DRXed α -Mg phase increased, with the grain size rising from 1.1 to $4.5 \mu\text{m}$ and the area fraction increasing from 44 to 94% . In contrast, when the extrusion temperature was elevated from 523 to 673 K , the grain size of the DRXed α -Mg phase increased from 1.3 to $2.9 \mu\text{m}$, whereas the area fraction remained relatively constant at approximately 70% .

Fig. 9 illustrates the effects of extrusion conditions on the mean KAM value in the extruded $\text{Mg}_{97.37}\text{Zn}_{1.88}\text{Y}_{0.75}$ alloys. With increasing extrusion ratio, extrusion ram speed, and ex-

trusion temperature, the mean KAM value decreases, and the non-recrystallized grains exhibit a higher mean KAM value than the recrystallized grains.

4. Discussion

4.1. Effects of extrusion processing on thermal conductivity

In our previous study [31], by optimizing the heat treatment conditions of the $\text{Mg}_{97.37}\text{Zn}_{1.88}\text{Y}_{0.75}$ alloy, Zn and Y elements with negatively large mixing enthalpy largely com-

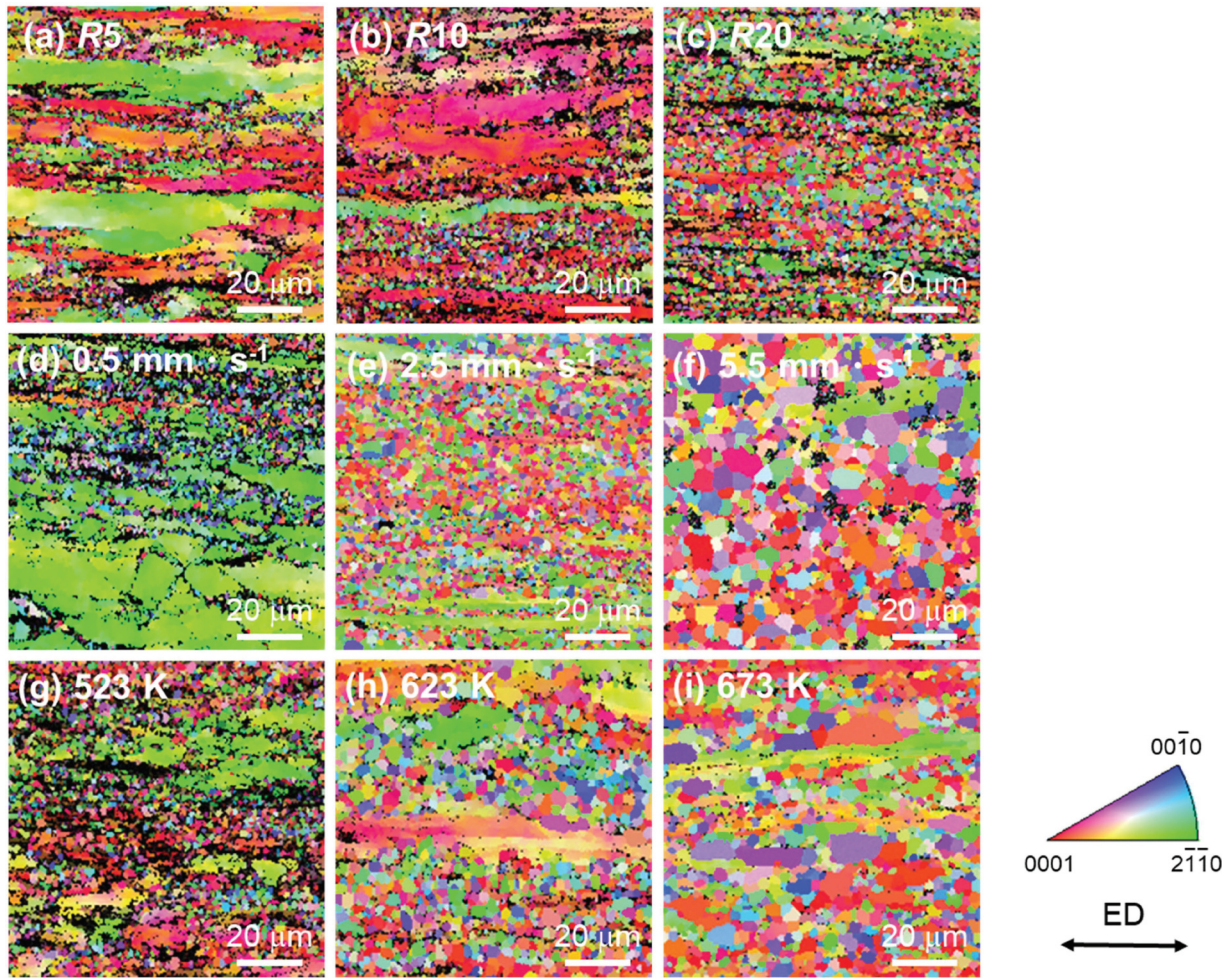


Fig. 7. IPF maps showing effects of extrusion ratio [(a) R5, (b) R10, (c) R20], extrusion ram speed [(d) $0.5 \text{ mm} \cdot \text{s}^{-1}$, (e) $2.5 \text{ mm} \cdot \text{s}^{-1}$, (f) $5.5 \text{ mm} \cdot \text{s}^{-1}$], and extrusion temperature [(g) 523 K, (h) 623 K, (i) 673 K] on extruded $\text{Mg}_{97.37}\text{Zn}_{1.88}\text{Y}_{0.75}$ (at%) alloys.

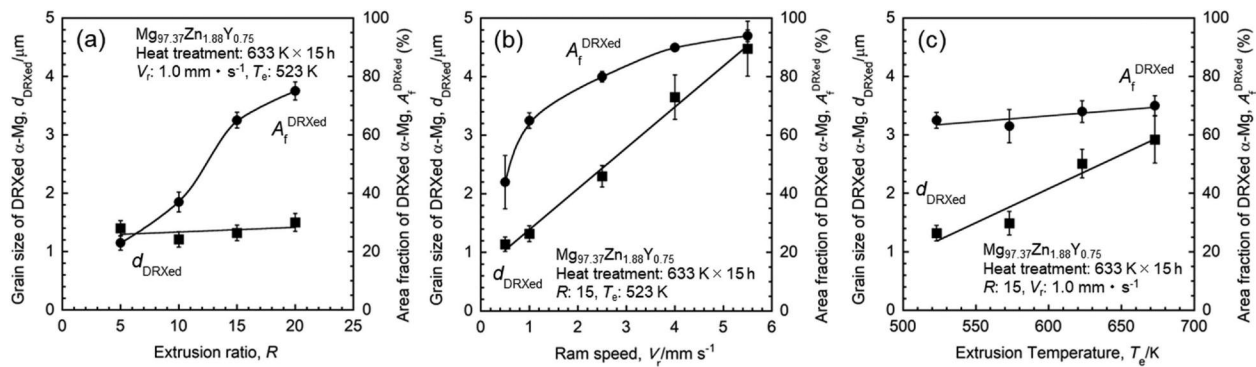


Fig. 8. Effects of (a) extrusion ratio (R), (b) extrusion ram speed (V_r), and (c) extrusion temperature (T_e) on the area fraction and grain size of dynamic recrystallized α -Mg in extruded $\text{Mg}_{97.37}\text{Zn}_{1.88}\text{Y}_{0.75}$ (at%) alloys.

bined to form the W-phase and precipitated, thereby purifying the matrix. As a result, the thermal conductivity of the alloy was significantly enhanced, exhibiting an ultrahigh value of $141 \text{ W m}^{-1}\text{K}^{-1}$. To enhance the mechanical properties of this high-thermal-conductivity alloy, we subjected it to extrusion processing. As illustrated in Fig. 3, although extrusion significantly enhanced the mechanical properties of the

$\text{Mg}_{97.37}\text{Zn}_{1.88}\text{Y}_{0.75}$ alloy, it concurrently caused a decrease in thermal conductivity. From the results in Figs. 1 and 3(b), we can find that in this study, regardless of extrusion conditions, the thermal conductivity of the $\text{Mg}_{97.37}\text{Zn}_{1.88}\text{Y}_{0.75}$ alloy decreased by $10 \pm 1 \text{ W m}^{-1}\text{K}^{-1}$ after extrusion processing. This phenomenon has also been observed in Mg–Al–Zn, Mg–Zn–Mn, and Mg–Al–Ca alloys. After extrusion processing, the

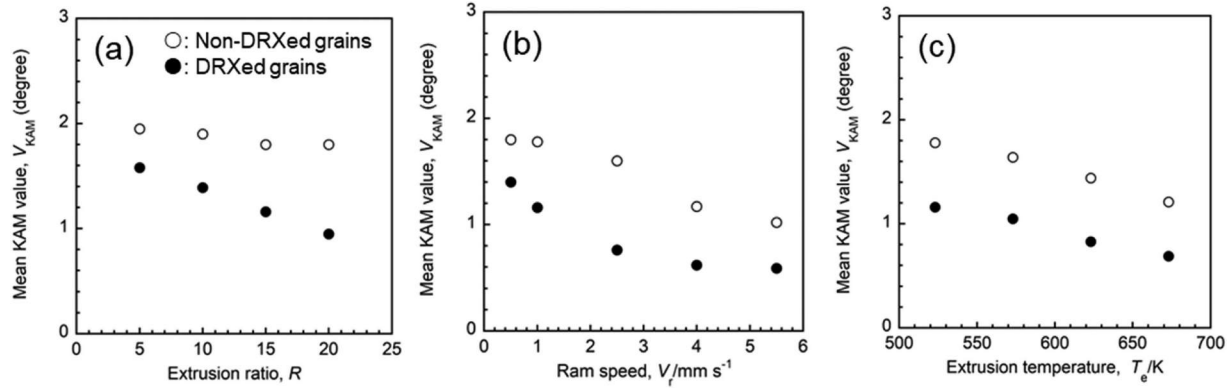


Fig. 9. Effects of (a) extrusion ratio (R), (b) extrusion ram speed (V_r), and (c) extrusion temperature (T_e) on the mean KAM value of dynamic recrystallized (DRXed) and non-recrystallized (Non-DRXed) α -Mg in extruded $Mg_{97.37}Zn_{1.88}Y_{0.75}$ (at%) alloys.

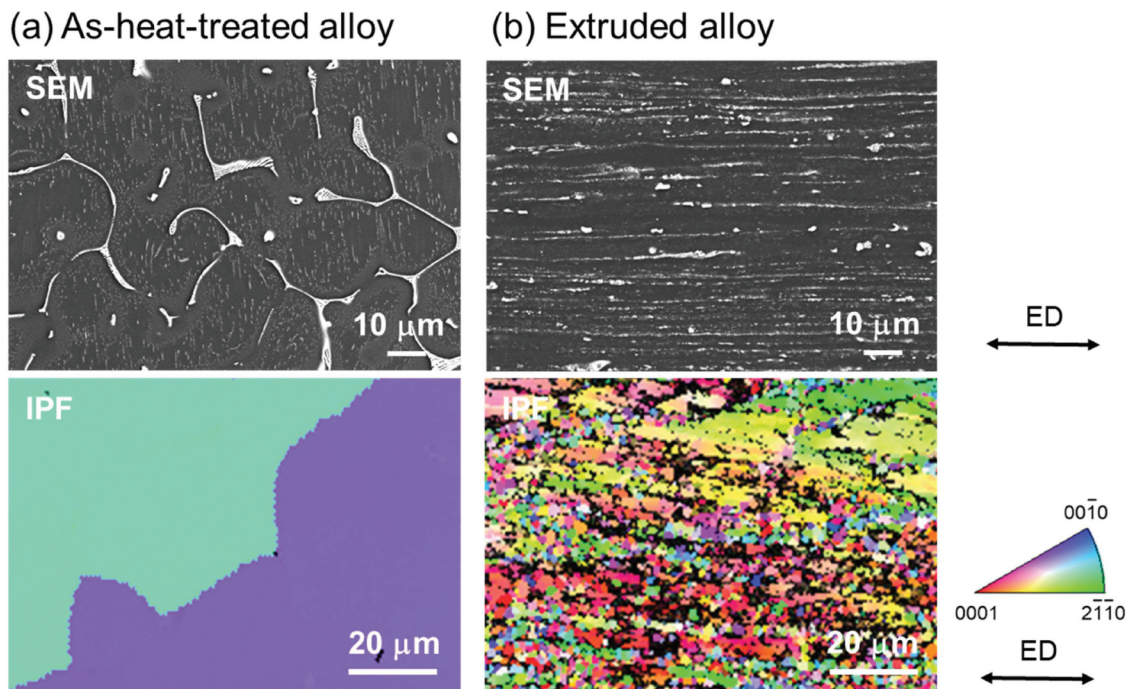


Fig. 10. SEM images and IPF map showing the microstructure of the $Mg_{97.37}Zn_{1.88}Y_{0.75}$ (at%) alloy in (a) heat-treated and (b) extruded states. The extruded material was produced under optimum extrusion conditions. The corresponding extrusion ratio (R), extrusion ram speed (V_r), and extrusion temperature (T_e) were 15, 1.0 mm·s⁻¹, and 523 K, respectively.

thermal conductivity of these alloys also decreased by 1–11 Wm⁻¹K⁻¹ [9,27,34–36]. Fig. 10 shows the microstructures of the heat-treated and extruded $Mg_{97.37}Zn_{1.88}Y_{0.75}$ alloys. Microstructural comparison revealed that after the extrusion processing, the coarse compounds were fragmented and, together with the fine precipitates, highly dispersed along the extrusion direction in a band-like distribution. This increased the probability of free electrons and phonons being scattered by the compounds. Simultaneously, the extrusion processing caused the dynamic recrystallization of coarse grains, decreasing the grain size from approximately 1000 μm to about 1 μm , and produced highly strained non-recrystallized grains with a pronounced texture. This led to an increase in the density of defects such as grain boundaries and dislocations, generating additional scattering sources for electrons and phonons.

Table 1

Solute concentration in α -matrix of heat-treated (HT: 633 K $\times$ 15 h, AC) and extruded (Ex: $R = 15$, $V_r = 1.0$ mm·s⁻¹, $T_e = 523$ –673 K) $Mg_{97.37}Zn_{1.88}Y_{0.75}$ alloys.

State	Solute concentration in α -Mg (at%)	
	Y	Zn
HT	0.04	0.70
Ex	0.07	0.71

In Table 1, the solute concentrations in the α -Mg matrix of the heat-treated and extruded alloys are compared. The results indicate that the solute concentration remains nearly unchanged after extrusion. Fig. 11 shows the correlations of the mean KAM value and the thermal conductivity of the heat-

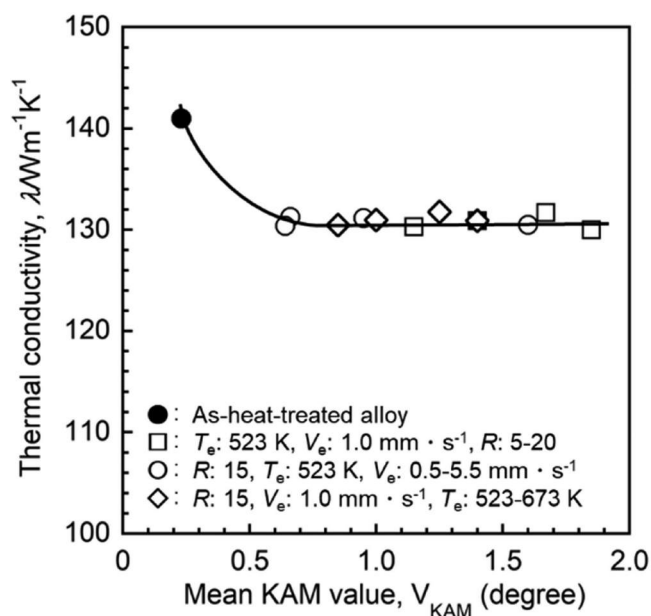


Fig. 11. Dependence of thermal conductivity on mean KAM value of heat-treated and extruded $\text{Mg}_{97.37}\text{Zn}_{1.88}\text{Y}_{0.75}$ (at%) alloys.

treated and extruded $\text{Mg}_{97.37}\text{Zn}_{1.88}\text{Y}_{0.75}$ alloys under different extrusion conditions. Compared with the heat-treated alloy, the extruded alloys exhibited a higher KAM value and a lower thermal conductivity. In other words, it is considered that the reduction in thermal conductivity occurred because the extrusion processing introduced various defects that affect thermal conductivity, such as grain boundaries, twin boundaries, dislocations, point defects, and nanovacancies. Since defects have a considerably smaller impact on thermal conductivity than solute atoms [23,24], the alloy could retain its high thermal conductivity even after extrusion.

After the extrusion processing, owing to the alignment of compound bands along the extrusion direction and the formation of texture, as shown in Fig. 1, the extruded Mg alloys typically exhibit anisotropic thermal conductivity, with the transverse direction (TD) 6–9 $\text{Wm}^{-1}\text{K}^{-1}$ higher than the extrusion direction (ED). This phenomenon has been reported in several studies [34–37]. However, compared with Mg–Al–Ca–Mn alloys [9], the alloys in the present study show slightly greater anisotropy, which is presumed to be due to the addition of the rare-earth element Y, leading to the formation of a certain rare-earth texture within the recrystallized grains [38,39].

The thermal conductivity of metals is primarily governed by the movement of free electrons and lattice vibrations (phonons), which can be scattered by various microstructural obstacles, such as impurities, grain boundaries, and phase boundaries, leading to a reduction in alloy thermal conductivity. Fig. 12 shows the electrical and thermal conductivities of the as-cast, heat-treated, and extruded $\text{Mg}_{97.37}\text{Zn}_{1.88}\text{Y}_{0.75}$ alloys. A linear relationship between thermal conductivity and electrical conductivity is observed, consistent with the Wiedemann–Franz law ($\lambda_e/\sigma = LT$). By comparing the cal-

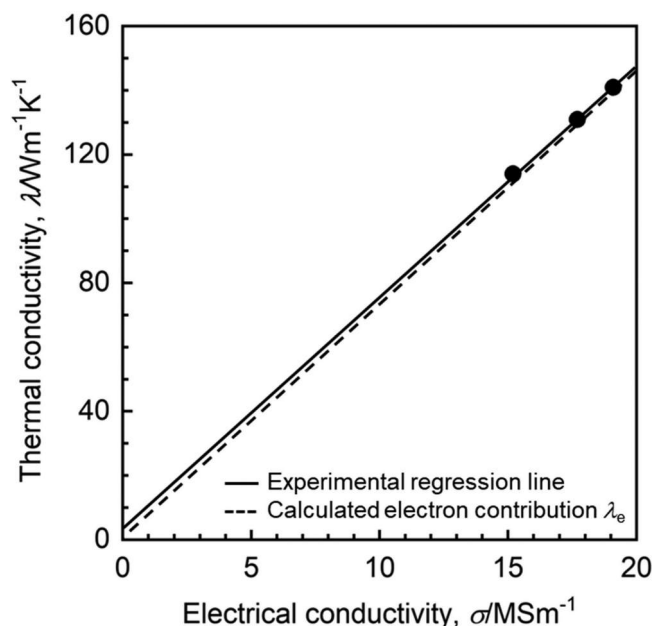


Fig. 12. Relationship between the thermal and electrical conductivities of the as-cast, heat-treated (633 K $\times$ 15 h, AC), and extruded ($R = 15$, $V_r = 1.0$ $\text{mm}\cdot\text{s}^{-1}$, $T_e = 523$ K) $\text{Mg}_{97.37}\text{Zn}_{1.88}\text{Y}_{0.75}$ (at%) alloys.

culated λ_e – σ line with the experimentally measured λ – σ line, we estimated that electrons contributed approximately 98 % of the total thermal conductivity in this alloy. This contribution is likely attributable to the extensive formation and precipitation of compounds, which effectively purify the matrix, thereby rendering the effect of phonon-impurity scattering negligible.

4.2. Effects of extrusion processing on mechanical properties of $\text{Mg}_{97.37}\text{Zn}_{1.88}\text{Y}_{0.75}$ alloy

Figs. 13(a), (b), (c), and (d) respectively present the correlations of the dispersion of compound bands, the particle size of compounds, the area fraction of the dynamic recrystallized α -Mg, and the grain size of dynamic recrystallized α -Mg and the tensile yield strength of the extruded $\text{Mg}_{97.37}\text{Zn}_{1.88}\text{Y}_{0.75}$ alloys (633 K $\times$ 15 h + Ex) under different extrusion conditions. The tensile yield strength of the extruded $\text{Mg}_{97.37}\text{Zn}_{1.88}\text{Y}_{0.75}$ alloys under different extrusion conditions is significantly positively correlated to the decline in the grain size of the dynamic recrystallized α -Mg. In accordance with the Hall Petch law, refining grains can produce more grain boundaries, which impose a higher resistance on the movement of dislocations, consequently resulting in an increase in the yield strength of the alloy. The Hall–Petch law can be expressed as [40]:

$$\sigma_{gb} = \sigma_0 + kd^{-\frac{1}{2}}, \quad (4.1)$$

where σ_{gb} is the grain boundary strengthening, σ_0 is the intrinsic strength of the material, k is the slope (approximately 190 $\text{MPa}\cdot\mu\text{m}^{1/2}$, as shown in Fig. 13(d)), and d means the average grain size. Thus, it can be calculated

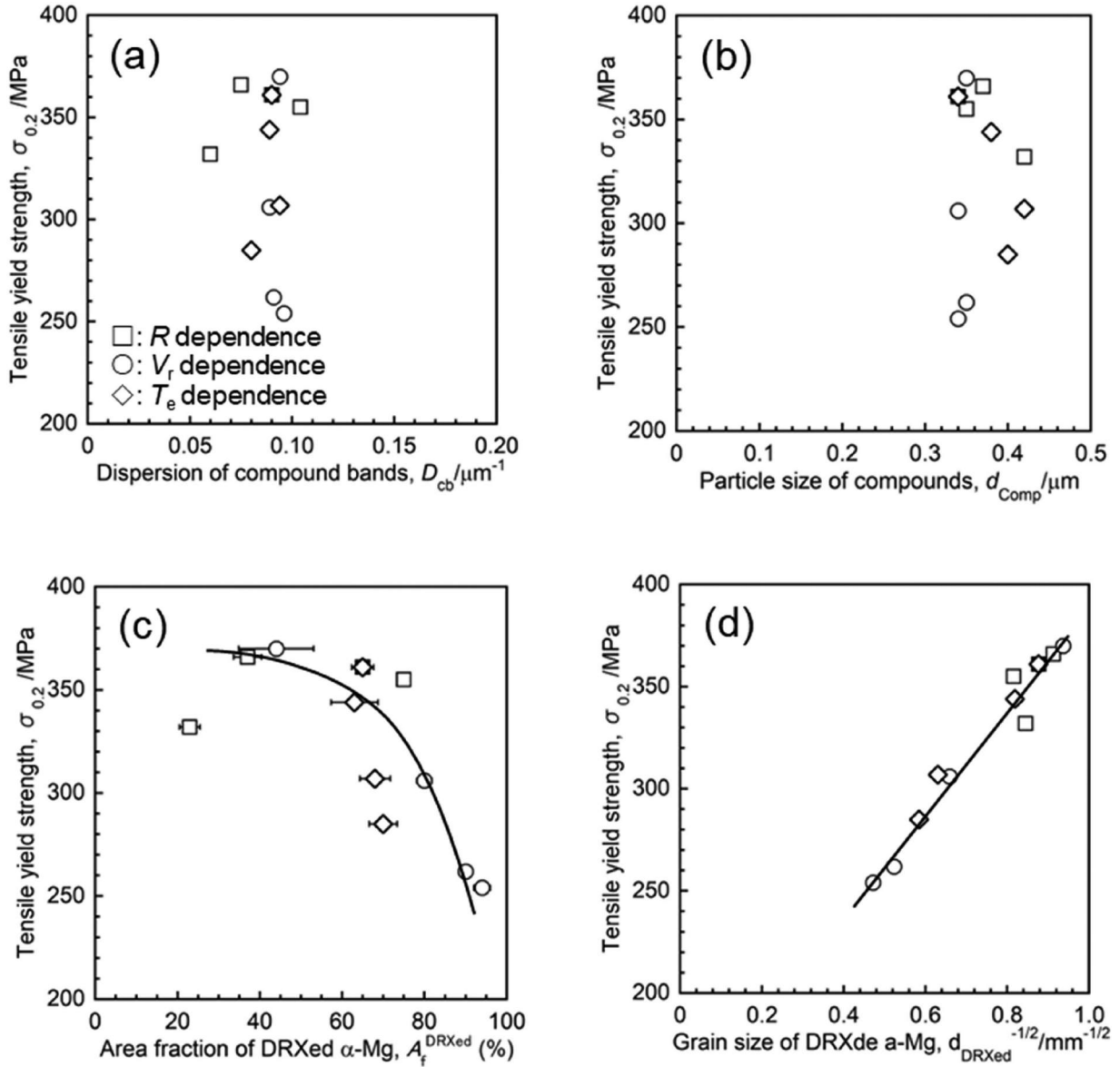


Fig. 13. The correlations of (a) the dispersion of compound bands, (b) the particle size of compounds, (c) the area fraction of the dynamic recrystallized α -Mg, and (d) the grain size of the dynamic recrystallized α -Mg with the tensile yield strength of the extruded $\text{Mg}_{97.37}\text{Zn}_{1.88}\text{Y}_{0.75}$ (at%) alloys ($633\text{ K} \times 15\text{ h} + \text{Ex}$) under different extrusion conditions.

that in this experiment, the grain boundary strengthening caused by the fine recrystallized grains in the extruded samples is approximately 37–116 MPa. In addition, there is also a negative correlation between the area fraction of the dynamic recrystallized α -Mg and the yield strength of the alloy. This is because the fewer α -Mg grains that achieve dynamic recrystallization, the more non-recrystallized grains will be stretched and deformed along the extrusion direction. As is well known, compared with dynamically recrystallized grains, non-recrystallized worked grains have higher dislocation density. Therefore, more non-recrystallized areas can also produce greater dislocation strengthening (σ_{dis}) in the alloy. The formula of σ_{dis} can be expressed as [40,41]:

$$\sigma_{\text{dis}} = A_f M \alpha G b \sqrt{\rho}, \quad (4.2)$$

where A_f represents the area fraction of non-recrystallized grains, M represents the Taylor factor of non-recrystallized grains (taken as 3.5 for Mg [41]), α is a constant (taken as 0.2 for Mg [41]), G is the shear modulus (taken as 16.6 GPa for Mg [42]), b is the Burger's vector (taken as 3.21×10^{-10} m for Mg [43]), and ρ means the dislocation density of non-recrystallized grains, which is calculated as [40,43,44]:

$$\rho = \frac{2\theta_{\text{KAM}}}{\mu b}, \quad (4.3)$$

where θ_{KAM} means the average KAM value of the non-recrystallized grains, μ is the EBSD scanning step size (0.6 μm), and b is the Burger's vector. From these formulas, σ_{dis} can be calculated. Fig. 14 shows the relationship between the area fraction of non-recrystallized grains (A_f) and the dislo-

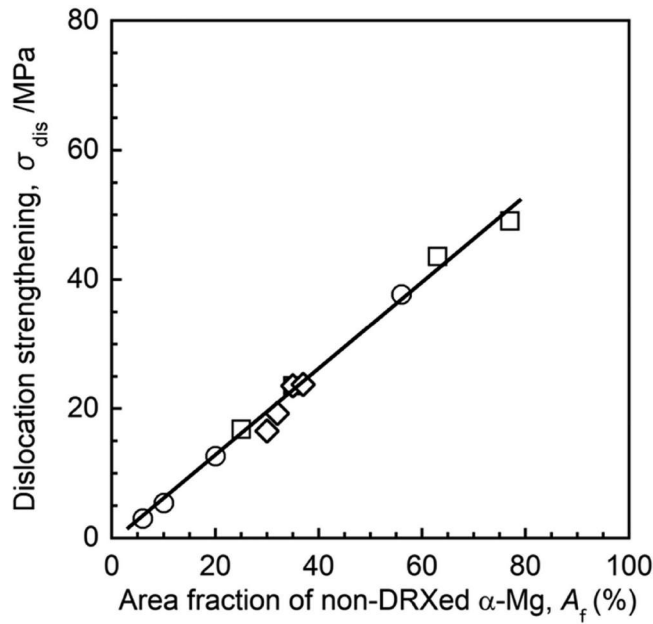


Fig. 14. Relationship between the area fraction of non-recrystallized grains ($A_f^{\text{non-DRXed}}$) and the dislocation strengthening (σ_{dis}) of the $\text{Mg}_{97.37}\text{Zn}_{1.88}\text{Y}_{0.75}$ (at%) alloys.

cation strengthening (σ_{dis}). Dislocation strengthening shows a significant positive correlation with the area fraction of non-recrystallized grains. The second phase plays a non-negligible role in achieving the high strength of the alloy [32,33]. In this study, the volume fraction of the compounds remained constant, and changes in the extrusion conditions did not alter the particle size or the dispersion of compound bands. Therefore, in the extruded alloys in this study, no significant relationship was observed between the yield strength and the particle size or the dispersion of compound bands.

Figs. 15(a), (b), (c), and (d) respectively present the correlations of the dispersion of compound bands, the particle size of compounds, the area fraction of the dynamic recrystallized α -Mg, and the grain size of the dynamic recrystallized α -Mg with the elongation of the extruded $\text{Mg}_{97.37}\text{Zn}_{1.88}\text{Y}_{0.75}$ alloys ($633 \text{ K} \times 15 \text{ h} + \text{Ex}$) under different extrusion conditions. A significant positive correlation is observed between the elongation of the alloys and the area fraction of the dynamically recrystallized α -Mg. The dynamic recrystallized α -Mg has a more random crystal orientation, and the increase in the area fraction of the dynamic recrystallized α -Mg can effectively increase the Schmid factor, making sliding easier. Similarly, to tensile strength, in this study, no significant relationship was found between ductility and the particle size and dispersion of compounds.

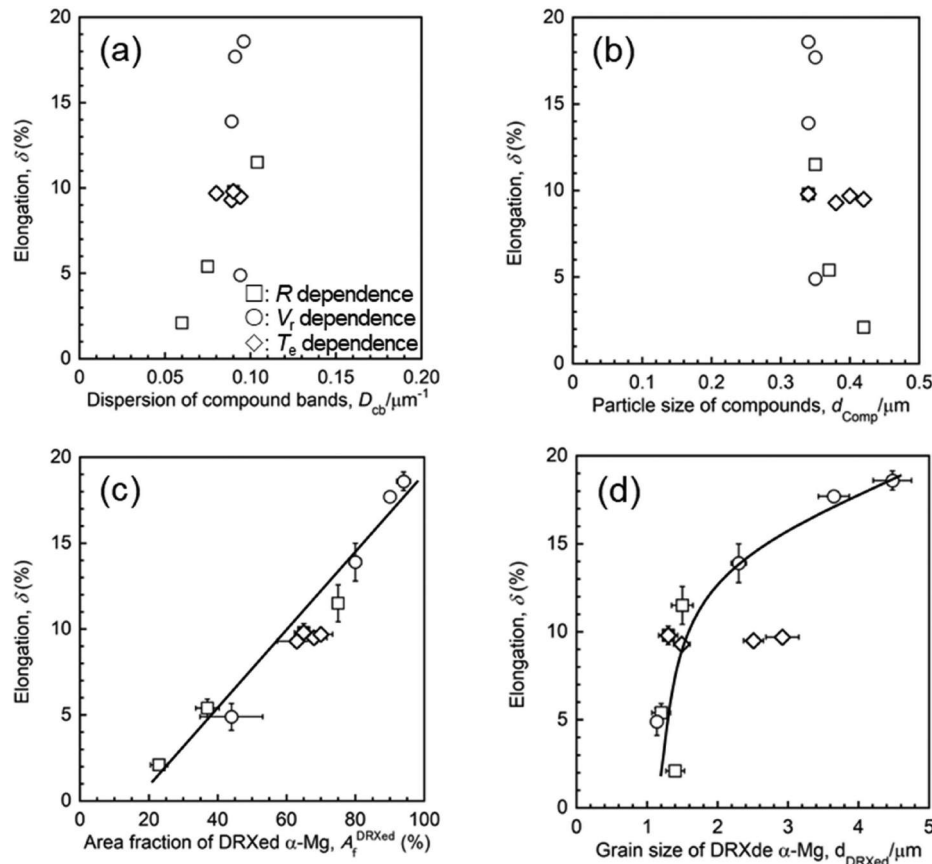


Fig. 15. The correlations of (a) the dispersion of compound bands, (b) the particle size of compounds, (c) the area fraction of the dynamic recrystallized α -Mg, and (d) the grain size of the dynamic recrystallized α -Mg with the elongation of the extruded $\text{Mg}_{97.37}\text{Zn}_{1.88}\text{Y}_{0.75}$ (at%) alloys ($633 \text{ K} \times 15 \text{ h} + \text{Ex}$) under different extrusion conditions.

Table 2

Properties of the developed Mg_{97.37}Zn_{1.88}Y_{0.75} (at%) alloys compared with those of existing high-thermal-conductivity and high-strength Mg alloys.

Alloy composition (at%)	Process conditions				Properties			
	Heat treatment before extrusion	Extrusion			Heat treatment after extrusion	$\sigma_{0.2}$ (MPa)	δ (%)	λ_{ED} (W m ⁻¹ K ⁻¹)
		R	V_f (mm/s)	T_e (K)				
Mg _{97.37} Zn _{1.88} Y _{0.75}	633 K × 15 h, AC	10	1.0	523	As extruded	366 ± 2	5.4 ± 0.5	132 ± 0.2
	633 K × 15 h, AC	15	1.0	523	As extruded	361 ± 2	9.8 ± 0.4	131 ± 0.1
	633 K × 15 h, AC	20	1.0	523	As extruded	355 ± 3	11.5 ± 1.0	130 ± 0.3
	633 K × 15 h, AC	15	1.0	573	As extruded	350 ± 4	9.3 ± 0.3	132 ± 0.2
	633 K × 15 h, AC	15	1.0	623	As extruded	307 ± 2	9.5 ± 0.3	131 ± 0.2
	633 K × 15 h, AC	15	2.5	523	As extruded	306 ± 5	13.9 ± 1.0	131 ± 0.1
	633 K × 15 h, AC	15	1.0	673	As extruded	285 ± 2	9.7 ± 0.3	131 ± 0.1
	633 K × 15 h, AC	15	4.0	673	As extruded	262 ± 4	17.7 ± 0.3	131 ± 0.2
Mg _{92.98} Al _{4.5} Ca _{2.5} Mn _{0.02}	673 K × 2.5 h, CFC	20	2.5	473	As extruded	397 ± 3	5.5 ± 0.5	124 ± 0.2
	673 K × 2.5 h, CFC	20	2.5	493	As extruded	384 ± 2	6.0 ± 0.2	123 ± 0.3
	673 K × 2.5 h, CFC	20	2.5	523	As extruded	376 ± 2	7.3 ± 0.2	123 ± 0.1
	673 K × 2.5 h, CFC	20	2.5	548	As extruded	358 ± 2	7.6 ± 0.1	121 ± 0.1
	673 K × 2.5 h, CFC	20	2.5	573	As extruded	356 ± 2	7.7 ± 0.2	120 ± 0.2
	673 K × 2.5 h, CFC	20	2.5	598	As extruded	358 ± 8	7.4 ± 0.4	125 ± 0.3
	673 K × 2.5 h, CFC	20	2.5	623	As extruded	333 ± 2	8.1 ± 1.3	120 ± 0.1
	673 K × 2.5 h, CFC	20	2.5	648	As extruded	326 ± 2	8.4 ± 0.1	119 ± 0.3
	673 K × 2.5 h, CFC	20	2.5	673	As extruded	313 ± 1	9.7 ± 0.7	119 ± 0.2
	673 K × 2.5 h, CFC	20	4.0	473	As extruded	377 ± 9	6.3 ± 0.8	124 ± 0.1
	673 K × 2.5 h, CFC	20	5.5	473	As extruded	352 ± 1	7.6 ± 0.1	120 ± 0.2
	673 K × 2.5 h, FC	20	2.5	623	As extruded	337 ± 4	5.2 ± 0.5	118 ± 0.7
Mg _{93.97} Al ₄ Ca ₂ Mn _{0.03}	673 K × 2.5 h, WC	20	2.5	623	As extruded	318 ± 4	7.4 ± 0.4	119 ± 0.1
Mg _{97.61} Zn _{1.93} Mn _{0.46}	643 K × 12 h	16	UNK	623	448 K × 36 h	268	14.8	122
	643 K × 12 h	16	UNK	623	673 K × 1.5 h, 448 K × 20 h	310	12.5	117
Mg _{96.38} Zn _{3.15} Mn _{0.47}	643 K × 12 h	16	UNK	623	448 K × 28 h	288	15.3	120
	643 K × 12 h	16	UNK	623	673 K × 1.5 h, 448 K × 28 h	354	12	115
Mg _{98.7} Zn _{1.14} Zr _{0.16}	UNK	UNK	UNK	UNK	As extruded	255	7	126
Mg _{97.76} Zn _{2.12} Zr _{0.12}	UNK	UNK	UNK	UNK	As extruded	260	11	117
Mg _{98.55} Zn _{0.96} Sm _{0.24} Zr _{0.25}	573 K	28	3.3	623	As extruded	254	11.3	124
Mg _{99.75} Mn _{0.22} Ce _{0.03}	693 K × 12 h	25	2.0	673	As extruded	293	7.6	137
Mg _{99.73} Mn _{0.22} Ce _{0.05}	693 K × 12 h	25	2.0	673	As extruded	196	9.6	140
Mg _{99.57} Mn _{0.4} Ce _{0.03}	773 K × 24 h	12	0.1	423	As extruded	270	12.5	112
Mg _{99.51} Mn _{0.4} Ce _{0.09}	773 K × 24 h	12	0.1	423	As extruded	285	11.3	113
Mg _{99.44} Mn _{0.4} Ce _{0.16}	773 K × 24 h	12	0.1	423	As extruded	320	7.4	121
Mg _{94.99} Al _{3.81} La _{1.11} Mn _{0.09}	623 K × 15 min	12	0.1	623	As extruded	256	8.4	128

Alloys with a tensile yield strength of 250 MPa or higher, elongation above 5%, and thermal conductivity over 100 Wm⁻¹K⁻¹ are listed for comparison.

The above results show that refining dynamically recrystallized grains and increasing their area fraction are effective for simultaneously achieving high strength and ductility.

4.3. Placement of extruded Mg_{97.37}Zn_{1.88}Y_{0.75} alloys among Mg-based wrought alloys with high thermal conductivity and mechanical strength

To date, Mg–Al–Ca, Mg–Zn–Mn, Mg–Zn–Zr, Mg–Zn–Sm–Zr, and Mg–Mn–Ce alloys have been reported as Mg alloys with high thermal conductivity [9,27–30,35,36,45,46]. Fig. 16 shows comparison of the thermal conductivity and tensile yield strength of the extruded Mg_{97.37}Zn_{1.88}Y_{0.75} alloys developed in this work with those of available high-performance Mg alloys ($\lambda \geq 100$ Wm⁻¹K⁻¹, $\sigma_{0.2} \geq 250$ MPa, and $\delta \geq 5\%$). The detail date of these alloys is listed in Table 2. It is evident that the extruded Mg_{97.37}Zn_{1.88}Y_{0.75} alloys developed in this work demonstrate thermal conductivity and mechanical strength superior to those of most previously reported high-performance Mg alloys. Although the yield strength of the present alloy is slightly lower than that of

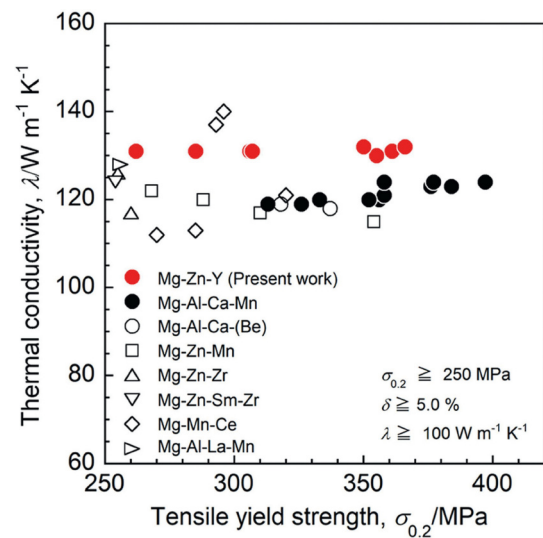


Fig. 16. Thermal conductivity and tensile yield strength of extruded Mg_{97.37}Zn_{1.88}Y_{0.75} (at%) alloys developed in this study compared with those of existing high-thermal-conductivity and high-strength Mg alloys ($\lambda \geq 100$ Wm⁻¹K⁻¹, $\sigma_{0.2} \geq 250$ MPa, and $\delta \geq 5\%$).

Mg–Al–Ca alloys, it demonstrates higher thermal conductivity and improved ductility, indicating a more favorable balance of properties for applications requiring both heat dissipation and mechanical performance.

5. Conclusions

To develop high-thermal-conductivity and high-strength Mg–Zn–Y alloys, the effects of different extrusion conditions on the extruded $\text{Mg}_{97.37}\text{Zn}_{1.88}\text{Y}_{0.75}$ alloy were investigated. The main results obtained are as follows:

- (1) By extruding the high-thermal-conductivity $\text{Mg}_{97.37}\text{Zn}_{1.88}\text{Y}_{0.75}$ alloy under optimized processing conditions, we successfully realized a multifunctional Mg–Zn–Y alloy with an excellent balance of thermal conductivity, strength, and ductility. The optimal extrusion conditions were determined to be $R = 15$, $V_r = 1.0 \text{ mm}\cdot\text{s}^{-1}$, and $T_e = 523 \text{ K}$. Under these conditions, the alloy exhibited a high thermal conductivity of $131 \text{ Wm}^{-1}\text{K}^{-1}$, a tensile yield strength of 361 MPa , and an elongation of 9.7% .
- (2) Regardless of the extrusion conditions, the extrusion processing caused the thermal conductivity to decline moderately by about $10 \pm 1 \text{ Wm}^{-1}\text{K}^{-1}$.
- (3) The microstructural analysis of the extruded $\text{Mg}_{97.37}\text{Zn}_{1.88}\text{Y}_{0.75}$ alloys revealed that there was an inverse relationship between the tensile yield strength of the alloy and the grain size of the dynamic recrystallized α -Mg. In contrast, the elongation was found to be positively correlated with the area fraction of the dynamic recrystallized α -Mg. These findings suggest that, in order to achieve a favorable combination of high strength and ductility, it is essential to both refine the dynamic recrystallized grain structure and maximize its area fraction.
- (4) The achievement of high strength in the high-thermal-conductivity $\text{Mg}_{97.37}\text{Zn}_{1.88}\text{Y}_{0.75}$ alloy demonstrates that, by optimizing the composition, heat treatment conditions, and extrusion conditions of the Mg alloy with a large negative mixing enthalpy of the added element pair, we may simultaneously attain high thermal conductivity, high strength, and excellent ductility in Mg alloys.

Declaration of competing interest

Professor Yoshihito Kawamura is an associate editor of the Journal of Magnesium and Alloys but was not involved in the editorial review or the decision to publish this article. All authors declare that there are no competing interests.

CRedit authorship contribution statement

Yunsheng Wang: Writing – original draft, Project administration, Methodology, Investigation, Formal analysis, Data

curation, Conceptualization. **Shin-ichi Inoue:** Writing – review & editing, Validation, Supervision, Project administration, Methodology, Investigation, Formal analysis, Conceptualization. **Yoshihito Kawamura:** Writing – review & editing, Validation, Supervision, Resources, Project administration, Methodology, Investigation, Formal analysis, Conceptualization.

Acknowledgments

This research was conducted with the support of the JSPS Grant-in-Aid for Scientific Research (A) (JP20H00312 and JP25H00807), the ATLA Security Science and Technology Research Promotion Program (JPJ004596), and JST SICORP V4 (JPMJSC2019).

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