第32卷第12期 中国有色金属学报 2022年 12月
Volume 32 Number 12 The Chinese Journal of Nonferrous Metals December 2022
DOI: 10.11817/j.ysxb.1004.0609.2021-42622

镁-稀土合金塑性变形技术研究进展

崔 磊[1]，唐昌平[1,2]，李 权[3]，刘 筱[1]

(1. 湖南科技大学 材料科学与工程学院，高功效轻合金构件成形技术及耐损伤性能评价
湖南省工程研究中心，高温耐磨材料及制备技术湖南省国防科技重点实验室，
新能源储存与转换先进材料湖南省重点实验室，湘潭 411201；
2. 华南理工大学 广东省精密装备与制造技术重点实验室，广州 510641；
3. 重庆市科学技术研究院，重庆 401123)

摘 要：综述了镁-稀土合金在挤压、轧制和大塑性变形技术方面的研究进展，介绍变形温度、变形速度、挤压比等对挤压材组织和力学性能的影响，发现通过变形工艺调控获得细晶或形成双峰分布晶粒可提高合金的力学性能。概述了镁-稀土合金轧制变形的研究进展，发现通过工艺调控形成双峰分布晶粒或引入层错可制备超高强镁合金，总结了等通道转角挤压、高压扭转和多向锻造对合金组织和力学性能的影响，发现大塑性变形技术尤其是高压扭转技术是制备纳米级超细晶的有效方法，但大塑性变形技术的工艺相较于挤压、轧制变形更复杂，成本更高，且制备的样品尺寸往往较小。最后，对镁-稀土合金塑性变形技术的发展方向提出了建议。

关键词：镁-稀土合金；挤压；轧制；大塑性变形；力学性能

文章编号：1004-0609(2022)-12-3632-17 **中图分类号**：TG306 **文献标志码**：A

引文格式：崔 磊, 唐昌平, 李 权, 等. 镁-稀土合金塑性变形技术研究进展[J]. 中国有色金属学报, 2022, 32(12): 3632−3648. DOI: 10.11817/j.ysxb.1004.0609.2021-42622

CUI Lei, TANG Chang-ping, LI Quan, et al. Research progress of plastic deformation technology of magnesium-rare earth alloy[J]. The Chinese Journal of Nonferrous Metals, 2022, 32(12): 3632−3648. DOI: 10.11817/j.ysxb.1004.0609.2021-42622

随着能源供应问题与环境问题日益突出，航空航天、交通运输等领域对构件轻量化的需求越来越迫切[1−2]。镁具有密度低、比强度高、比刚度高、阻尼减振性能好等优点，在构件轻量化制造中具有天然的优势，发展前景广阔[3−6]。然而，纯镁的强度低，往往难以直接应用于结构件中[7]。研究发现，向镁中添加稀土元素可显著提高合金的力学性能，常见的添加元素有Gd、Y等。Gd、Y元素在镁中均有较大的固溶度，且固溶度随温度的降低而降低，具备时效强化的条件[8]。经时效处理后，可形成盘状纳米级析出相，该相在棱柱面上呈三角分布，能有效阻碍基面位错的运动，显著提升合金的力学性能[9−10]。过去的几十年中，我国开发了一系列性能优异的镁-稀土合金材料，并在航空航天等领域获得了广泛的应用[11−13]。

相比铸造镁合金，变形镁合金的力学性能往往

基金项目：国家自然科学基金资助项目(52075167，51605159，52071139)；湖南省自然科学基金资助项目(2020JJ4307)；湖南省教育厅优秀青年项目(19B214)；广东省精密装备重点实验室开放课题(PEMT202103)；湖南省研究生科研创新项目(CX20211030)

收稿日期：2021-11-04；**修订日期**：2022-01-18

通信作者：唐昌平，副教授，博士；电话：18773260825；E-mail：tcpswnu@163.com

更为优异。例如，中南大学刘楚明团队通过挤压+旋转模锻的变形方式成功制备出抗拉强度高达710 MPa的Mg-Gd-Y-Zr合金[14]，其强度已达到了超高强铝合金的水平。铸造镁合金在塑性变形过程中，一些铸造缺陷可以被消除，粗大的第二相粒子也能被破碎，可有效降低裂纹源[10]，提高合金的力学性能。此外，镁合金为密排六方结构，在室温下仅基面滑移和锥面孪生容易启动，塑性较差[15−16]。当温度升高后，由于基面滑移、棱柱面滑移和锥面滑移临界分切应力之间的差异降低[16−17]，可启动更多的滑移系，有效协调合金的塑性变形，故镁合金的塑性变形通常在高温下进行。由于镁合金层错能低，在热变形过程中容易发生动态再结晶而使晶粒细化。同时，由于镁合金Hall-Petch关系中K值较铝合金的更高[18−20]，又可显著提升合金的晶界强化效果。计算表明[9]，除时效强化外，晶界强化是变形镁−稀土合金中最重要的强化机制。因此，在过去的二十几年中，挤压[21−28]、轧制[29]、大塑性变形[30−32]等塑性变形技术已广泛应用于镁−稀土合金中，制备了大量具有微米级、亚微米级甚至纳米级细晶的镁合金材料，获得了优异的力学性能。

本文作者综述了近年来镁−稀土合金塑性变形技术方面的研究进展，主要包括常规挤压、常规轧制以及等通道转角挤压(Equal channel angular pressing, ECAP)、高压扭转(High pressure torsion, HPT)和多向锻造(Multi-directional forging, MDF)等大塑性变形技术(Severe plastic deformation, SPD)，重点关注了塑性变形对合金微观组织和力学性能的影响，以期为高性能镁合金构件的制备提供参考。

1　常规挤压

挤压是工业上常用的塑性变形方式，变形区呈三向压应力状态，有利于合金塑性的发挥，特别适合于镁合金这类塑性较低的材料。近年来，镁合金的挤压变形技术发展迅速，一系列强度超过400 MPa的Mg-RE挤压变形镁合金被成功开发出来，部分合金性能如表1所示[21−27, 33−43]。

挤压变形时的温度是影响镁合金性能的重要因素。研究发现，随挤压温度升高，再结晶更容易发生，再结晶体积分数增大[36, 38, 42]。HUANG等[38]采用新型挤压模具针对开发的JDBM-2.1Nd (Mg-2.1Nd-0.2Zn-0.4Zr)合金在250 ℃、280 ℃和310 ℃和340 ℃进行挤压，发现再结晶体积分数与温度呈正相关关系，340 ℃发生了完全再结晶。随挤压温度升高，合金的屈服强度降低，250 ℃获得了541 MPa的超高屈服强度。对高成分合金而言，挤压温度对合金微观组织的影响更为复杂。当挤压温度太低时，虽然能获得细小的晶粒，但会导致合金发生动态析出，降低后续时效过程中的沉淀强化效果。高温挤压又会导致晶粒粗大，降低晶界强化效果[44]。基于此，近年来开发了差热挤压工艺(铸锭预热温度和模具预热温度不同)，铸锭在高温预热有助于限制析出，模具温度在低温预热，控制实际挤压温度，有助于细化晶粒和避免表面裂纹。RONG等[35]对Mg-15Gd-1Zn-0.4Zr合金的等温挤压和差热挤压进行了对比研究。结果发现，差热挤压样品在挤压过程中的降温比等温挤压的样品更快、晶粒明显更细小。此外，差热挤压还可以通过改变铸锭温度和模具温度，进而控制变形后的冷却速度来调节长周期有序(Long period stacking ordered, LPSO)结构的形成和数量，差热挤压样品中的LPSO相只占3%，远低于等温挤压样品的25%。同时，与等温挤压相比，差热挤压可以采用更高的铸锭温度，产生更强的沉淀强化能力，时效后，差热挤压样品的抗拉强度相较于等温挤压样品提高了41 MPa，屈服强度提高了42 MPa。

挤压条件对合金的再结晶晶粒尺寸有显著的影响。一般来说，Zener−Hollomon (Z)参数与动态再结晶晶粒尺寸(d_{DRX})存在如下关系：

$$d_{\mathrm{DRX}}=AZ^{-n} \tag{1}$$

其中，

$$Z=\dot{\varepsilon}\exp\left(\frac{Q}{RT}\right) \tag{2}$$

$$\dot{\varepsilon}=\frac{6D_{\mathrm{B}}^{2}v_{\mathrm{R}}\ln E_{\mathrm{R}}}{D_{\mathrm{B}}^{3}-D_{\mathrm{E}}^{3}} \tag{3}$$

式中：A是常数；n是幂指数；Q是激活能；R是摩尔气体常数；T是温度；$\dot{\varepsilon}$是应变速率；D_{B}是铸锭直径；D_{E}是挤压棒直径；v_{R}是挤压速度；E_{R}是挤压比[45]。因此，由式(1)~(3)的计算可知，合金的动态再结晶晶粒尺寸随着挤压速度的增大而减小。然而，在实际研究中发现，随挤压速度增大，动态再结晶晶粒的尺寸反而越大[39−40, 46]。这主要是由于采

表1 强度超过400 MPa的挤压变形Mg-RE合金[21−27, 33−43]

Table 1 Extruded Mg-RE alloy with strength exceeding 400 MPa[21−27, 33−43]

Alloy	Processing condition		UTS/MPa	YS/MPa	EL/%	Ref.
Mg-8Gd-4Y-Nd-Zr	(520 ℃, 12 h)+extruded at 400 ℃(11.3∶1)+ (solution treatments, 0.5 h)+aged at 225 ℃		419	321	2.6	[21]
Mg-5.5Gd-3.0Y-1.0Nd-1.0Zr	Extruded at 400 ℃(16∶1, 30 mm/min)+ aged at (225 ℃, 12 h)		490(UCS)	325	8.9	[22]
Mg-9.2Gd-4.4Y-1.0Zn-0.8Mn	(500 ℃, 8 h)+(525 ℃, 4 h)+(pre-aged at 225 ℃, 110 h)+extruded at 420℃(11∶1)		455	382	11	[23]
Mg-11.5Gd-4.5Y-0.3Zr	(520 ℃, 48 h)+extruded at 450 ℃ (20∶1, 0.1 mm/s)+aged at 200℃		483	423	2.2	[24]
Mg-9.16Gd-2.12Y-0.24La-0.23Er-0.11Ho-0.64Zn-0.43Zr	(500 ℃, 10 h)+extruded at 360 ℃ (7∶1, 0.1 mm/s)+aged at 200 ℃		490	481	2.1	[25]
Mg-12Gd-2Y-1Zn-Mn	(515 ℃, 10 h)+(495 ℃, 4 h)+extruded at 495 ℃ (6.25∶1, 0.5 mm/s)+aged at 200℃		509	427	5.1	[26]
Mg-8.3Gd-4.2Y-1.4Zn-1.1Mn	As extruded+aged at (200℃, (0−250 h))		538	−	13.1	[27]
Mg-8Gd-1.2Zn-0.5Zr-0.5Ce	Extruded at 360 ℃(7∶1)+aged at (200 ℃, 64 h)		461	458	5.5	[33]
Mg-3.5Sm-2Yb-0.6Zn-0.4Zr	(525 ℃, 2 h)+(515 ℃, 4 h)+extruded at 300 ℃ (6.5∶1, 0.3 mm/s)+aged at 200 ℃		451	449	4.9	[34]
Mg-15Gd-1Zn-0.4Zr	(500 ℃, 2 h)+(520 ℃, 12 h)+extruded (billet at 490 ℃, mold at 370 ℃, 9∶1)+aged at 200 ℃		461	380	2.7	[35]
Mg-8Gd-4Y-Mn-0.4Sc	(520 ℃, 18 h)+extruded at 400 ℃ (10:1, 0.1 mm/s)		406	337	10.8	[36]
Mg-11.5Gd-4.5Y-0.3Zr	(520 ℃, 48 h)+extruded at 450℃ (10:1, 0.1 mm/s)		412	346	7.5	[37]
Mg-2.1Nd-0.2Zn-0.4Zr	(540 ℃, 10 h)+extruded at 250 ℃ (9:1)	250 ℃	−	541	3.7	[38]
		280 ℃	−	462	4.9	
Mg-9Gd-3Y-1.5Zn-0.8Zr	(530 ℃, 5 h)+extruded at 400 ℃ (15:1)+aged at 200 ℃	0.1 m/min	446	371	2.1	[39]
		0.3 m/min	407	325	2.1	
Mg-13.25Y-4.87Zn-1.29Li	(500 ℃, 35 h)+extruded at 350 ℃ (25:1)	1 mm/s	479(UCS)	312	26	[40]
		0.1 mm/s	632(UCS)	430	20.2	
Mg-8Gd-4Y-Mn-0.4Sc	(520 ℃, 18 h)+extruded at 400 ℃	(10∶1)	406	337	10.8	[36]
		(22∶1)	425	352	10.6	
Mg-11.5Gd-4.5Y-1.5Zn-0.3Zr	(520 ℃, 48 h)+extruded at 450 ℃	(10∶1)	420	356	7.6	[37]
		(20∶1)	424	371	7.2	
Mg-Gd-Y-Zn-Al	(520 ℃, 14 h)+extruded at 420 ℃ (0.2−0.3 mm/s)+aged at 200 ℃	(10∶1)	417	295	6.5	[41]
		(22∶1)	440	328	7.0	
Mg-7.5Gd-2.5Y-3.5Zn-0.9Ca-0.4Zr	Extruded at 400 ℃(10∶1, 0.1 mm/s)		405	345	8.8	[42]
Mg-13Gd	(510 ℃, 24 h)+extruded at 320 ℃ (4∶1, 0.2 mm/s)+aged at 200 ℃		508	470	2.5	[43]

UTS: Ultimate tensile strength; YS: Yield strength; EL: Elongation; UCS: Ultimate compressive strength.

用高速挤压时，快速的塑性变形和摩擦会产生大量的热导致温度升高，使再结晶晶粒尺寸增加。LIU 等[40]比较研究了 $Mg_{89}Y_4Zn_2Li_5$ 合金在 1 mm/s 和 0.1 mm/s 挤压后的微观组织和力学性能，发现挤压速度的降低显著限制了动态再结晶晶粒的长大，1 mm/s 和 0.1 mm/s 挤压样品的晶粒尺寸分别为4.51 μm和1.15 μm。从力学性能来看，1 mm/s挤压的合金，抗压强度为479 MPa，屈服强度为312 MPa；而 0.1 mm/s 挤压的合金，抗压强度提高到 632 MPa，屈服强度提高到430 MPa。降低挤压速度显著提升了合金的压缩强度。因此，为获得优异的力学性能，镁合金通常采用低速挤压。LI等[43]对Mg-13Gd合金的研究表明，当采用0.2 mm/s的速度挤压时，铸锭的实际变形温度不会上升太多，有利于获得较细的动态再结晶晶粒。

挤压比是影响合金微观组织和力学性能的另一重要因素。研究发现，随挤压比增大，动态再结晶晶粒的体积分数增大，变形晶粒的体积分数减少。大的挤压比提高了应变量和析出相的数量，促进了动态再结晶，使晶粒细化[36−37, 42, 47]。此外，随着挤压比的增大，织构强度也降低，这是由具有强织构的变形晶粒体积分数降低所致[36−37]。因此，力学性能通常随挤压比的增大而得到改善，例如，YANG 等[36]制备的Mg-Gd-Y-Mn-Sc合金，挤压比为10时，合金的抗拉强度和屈服强度分别为406 MPa和337 MPa，而挤压比为20时合金的抗拉强度和屈服强度分别提高至425 MPa和352 MPa。然而，最近的研究表明，小挤压比也能获得优异的力学性能。LI 等[43]对Mg-13Gd合金以4的挤压比进行了挤压(见图1)，由于挤压比较小变形不足，获得了变形粗晶粒和再结晶细晶粒构成的双峰分布组织，变形晶粒的体积分数约85%。变形晶粒因其强烈的基面织构，产生了织构强化，改善了合金的强度，获得了388 MPa的抗拉强度。时效后形成了大量的纳米级析出相进一步提升了强度，获得了508 MPa的超高强度，合金的强化机制由通常的细晶强化和沉淀强化转变为织构强化和沉淀强化。

挤压变形后进行固溶处理对合金的微观组织和力学性能也具有重要的影响。TANG 等[21]对 Mg-8Gd-4Y-Nd-Zr合金在挤压后进一步进行了固溶处理，发现450 ℃和475 ℃固溶处理后，在挤压过程中形成的 $Mg_{5.05}RE$ 析出相部分溶入了基体中。而经500 ℃和520 ℃固溶后，析出相大量减少甚至完全溶入了基体，但由于温度过高发生了晶粒的长大，降低了合金的变形能力。同时，经T6处理后，析出相的存在改变了基面滑移和锥面孪生临界分切应力的相对大小，使孪生更容易发生，导致合金过早断裂，使强度降低。475 ℃固溶的样品时效后获得了较优的性能，抗拉强度为419 MPa。

由上述分析可知，挤压变形是制备超高强度Mg-RE合金的方法之一，调整变形工艺能改善挤压材的力学性能。低温挤压能显著抑制再结晶晶粒的长大，但较低的挤压温度容易导致合金开裂，故差热挤压是实现晶粒细化和避免开裂的最佳方式。低速挤压能控制铸锭的温升，获得细小的再结晶晶粒，但存在生产效率低的缺点。大挤压比能有效细化晶粒，而小挤压比也可通过形成双峰分布晶粒而使合金强化，但也分别存在能耗高和工艺适用范围窄的缺点。

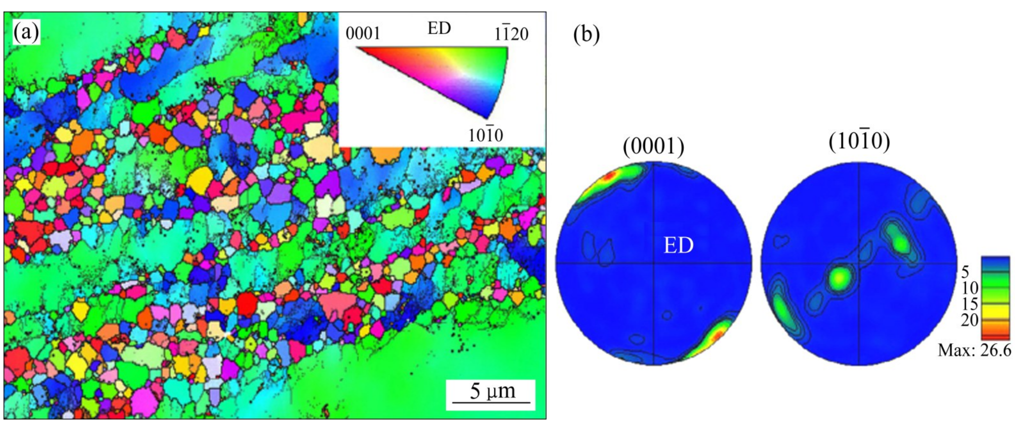

图1 挤压Mg-13Gd合金的双峰分布IPF图和极图[43]

Fig. 1 IPF map(a) and pole figure(b) of bimodal distribution microstructure of extruded Mg-13Gd alloy[43]

2 常规轧制

轧制变形是镁合金常用的常规变形方式之一，是制备镁合金板材很有效的变形方式。镁合金通过轧制变形可以不同程度地细化铸造镁合金中的粗大晶粒，提升合金的强度。近几年，研究人员通过控制变形条件制备了一系列强度大于400 MPa的Mg-RE合金(见表2[29, 48−52])。

镁合金轧制变形时的变形量对合金的微观组织和力学性能有显著影响。研究发现，镁合金轧制变形时，随着变形量的增大，合金的动态再结晶体积分数增大，晶粒尺寸显著细化[49, 53−55]。孪晶的演变也与变形量有关，研究发现，变形量较低时，合金内部形成大量的拉伸孪晶以及少量的压缩孪晶和二次孪晶[53, 55]。随着变形量的增大，拉伸孪晶的数量逐渐减少，而压缩孪晶和二次孪晶的数量增多，轧制变形量的增加有利于二次孪晶的形成[55]。

轧制变形时进行道次间的热处理对合金也有影响。先前的研究表明，退火处理会使合金发生软化[56]。然而刘栩东等[50]发现，Mg-10Gd-3Y-0.3Zr在轧制变形时引入了形变孪晶，低温退火处理具有形变孪晶的合金可以显著提高合金的抗压强度，这是因为退火时溶质原子对孪晶界起钉扎作用，因此具有更大的变形抗力[50]，最终获得了424 MPa的抗压强度。此外，轧制道次间的中间退火处理可以诱发再结晶[57]，轻微的变形都会使合金发生再结晶行为。

近年来，研究人员还通过形成双峰分布晶粒或引入层错制备了超高强度的轧制板材。XU等[52]采用大道次压下量(30%~45%)制备了Mg-8.2Gd-3.8Y-1.0Zn-0.4Zr轧制板材，形成了变形粗晶粒和再结晶细晶粒共存的双峰分布组织，获得了优异的力学性能，室温抗拉强度达到517 MPa，屈服强度达到426 MPa，伸长率为4.5%。JIAN等[48]发现Mg-

表2 强度超过400 MPa轧制变形后Mg-RE合金的力学性能[29, 48−52]

Table 2 Mechanical properties of rolled Mg-RE alloy with strength exceeding 400 MPa[29, 48−52]

Alloy	Processing condition	UTS/MPa	YS/MPa	EL/%	Ref.
Mg-11Gd-2Ag	(500 ℃, 24 h)+rolled at 470 ℃ (50%)+aged at 200 ℃	435	393	5.2	[29]
Mg-8.5Gd-2.3Y-1.8Ag-0.4Zr	(500 ℃, 12 h)+rolled at 450 ℃ (50%−88%)	600	575	5.2	[48]
Mg-13Gd-4Y-2Zn-0.6Zr	(520 ℃, 12 h)+rolled at 450 ℃ (89%, 94%)	415	301	9.91	[49]
Mg-10Gd-3Y-0.3Zr	(510 ℃, 6 h)+rolled at 450 ℃ (75%)	424	−	−	[50]
Mg-9Y-6Zn	Rolled at room temperature	−	460 (CYS)	−	[51]
Mg-8.2Gd-3.8Y-1.0Zn-0.4Zr	(510 ℃, 12 h)+rolled at 400 ℃ (96%)+aged at 200 ℃	517	426	4.5	[52]

UTS: Ultimate tensile strength; YS: Yield strength; EL: Elongation; CYS: Compressive yield strength.

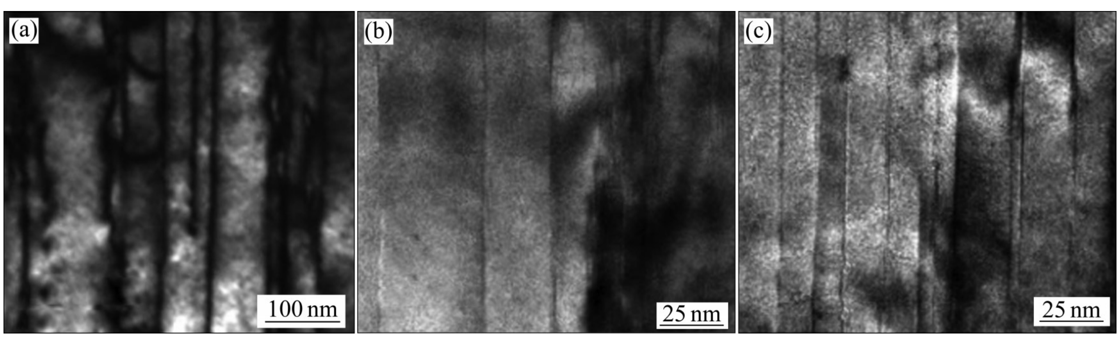

图2 Mg-8.5Gd-2.3Y-1.8Ag-0.4Zr合金不同变形量时的层错密度[48]

Fig. 2 Stacking fault density of Mg-8.5Gd-2.3Y-1.8Ag-0.4Zr alloy at different strains[48]: (a) 50%; (b) 70%; (c) 88%

8.5Gd-2.3Y-1.8Ag-0.4Zr合金在30%~50%的变形量之间存在一个过渡阶段，变形机制由位错滑移过渡为层错的协调作用。随着变形量从50%增加到88%，合金中的层错密度不断增大(见图2[48])。高密度的层错可有效阻碍位错的运动，从而提高合金的强度，最终使合金获得了600 MPa的超高抗拉强度，这也是迄今为止所报道的强度最高的镁合金轧制样品。

常规轧制制备超高力学性能镁−稀土合金板材的工艺关键在于采用大的道次变形量形成双峰分布晶粒，对铸锭质量和合金塑性变形能力要求极高，适用范围受到一定的限制。引入层错也是制备超高力学性能轧制板材的有效方法，但报道极少，可能需要在特定成分的合金中才能实现。

3　大塑性变形

3.1　等通道转角挤压

等通道角挤压(ECAP)技术往往显示出比传统变形工艺更为显著的晶粒细化效果，可以获得纳米级别的晶粒[58]，是目前发展最迅速的制备块状全致密超细晶(Ultra-fine grain, UFG)材料的大塑性变形(SPD)技术。已有大量研究表明，镁合金通过等通道转角挤压变形技术可以实现强度的提高以及塑性的增强[59−60]。近年来，研究人员对Mg-Y系、Mg-Gd-Ag、Mg-Gd-Y系进行了大量的研究，部分强度超过400 MPa的Mg-RE合金如表3所示[61−71]。

向Mg-Y合金中添加Zn元素后，在铸态合金中可形成18R-LPSO相，该相富Y和Zn元素，通常呈网络状结构[61−62, 72]或板条状结构[63]。在挤压过程中，18R-LPSO相将由初始态的板条状逐渐发生弯曲和扭折[61, 64]，进而产生裂纹并破碎[62−63, 72−73]。扭折的18R-LPSO相可通过粒子激发形核(Particle stimulated nucleation, PSN)机制促进动态再结晶而产生细晶强化，18R-LPSO相本身也可有效阻碍位错运动使合金强化。当破碎成颗粒后，细小的颗粒也可以强化合金[63]。然而，完全破碎的18R-LPSO相颗粒可能会降低合金表面膜的保护作用导致耐腐蚀性降低[72]。除此之外，在挤压过程中还会析出层状14H-LPSO相，该相可以阻碍晶界的移动，抑制晶粒长大[61]。合金成分对Mg-Y-Zn合金的力学性能具有显著的影响，随着Y元素和Zn元素含量的增加，合金的强度先增加后降低，当合金成分为Mg-4Y-2Zn时，经16道次挤压后，获得了优异的力学

表3　等通道转角挤压后强度大于400 MPa Mg-RE合金的力学性能[61−71]

Table 3　Mechanical properties of Mg-RE alloy with strength greater than 400 MPa after equal channel angular pressing[61−71]

Alloy	Processing condition	UTS/MPa	YS/MPa	EL/%	Ref.
Mg-4Y-2Zn	ECAP at 360 ℃ (5 mm/min,16 passes)	611 (UCS)	−	20.1	[61]
Mg-2Y-1Zn	ECAP at 350 ℃ (5 mm/min,16 passes)	446 (UCS)	−	16.5	[62]
Mg-8Y-4Zn	ECAP at 350 ℃ (5 mm/min,12 passes)	530 (UCS)	−	17.1	[63]
Mg-4Y-2Zn-0.04Sr	ECAP at 330 ℃ (2 mm/min, 4 passes)	408	−	5.1	[64]
Mg-10.6Gd-2Ag	(500 ℃, 24 h)+ECAP at 300 ℃ (3.5 mm/s, 8 passes)	408.4	319.5	18.2	[65]
	(500 ℃, 24 h)+ECAP at 350 ℃+aged at 200 ℃ (3 mm/s, 8 passes)	460.3	396.4	8.9	[66]
	(500 ℃, 24 h)+ECAP at 350 ℃+aged(8 passes)	502	409	8.8	[67]
	(500 ℃, 24 h)+ECAP at 350 ℃+ aged at 200℃ (3 mm/s, 8 passes)	460.3	396.4	8.9	[68]
Mg-10Gd-2Y-1.5Zn-0.5Zr	ECAP at 360 ℃ (5 mm/min, 8 passes)	518 (UCS)	263	21.6	[69]
Mg-10Gd-4Y-1.5Zn-0.5Zr	ECAP at 360 ℃ (5 mm/min, 16 passes)	548.2 (UCS)	300	19.1	[70]
Mg-10Gd-6Y-1.5Zn-0.5Zr	ECAP at 360 ℃ (5 mm/min, 8 passes)	537 (UCS)	361	17.0	[71]

UTS: Ultimate tensile strength; YS: Yield strength; EL: Elongation; UCS: Ultimate compressive strength.

性能，其抗压强度可达到611 MPa，最大压缩率可达到20.1%[61]。通过添加Sr元素[64]，对Mg-Y-Zn合金进行成分优化后，也获得了良好的效果。Mg-4Y-2Zn-0.04Sr合金在330 ℃进行4道次ECAP后，获得了408 MPa的抗拉强度。

Mg-Gd系合金具有显著的时效硬化效果。研究表明，Mg-Gd系合金通常在棱柱面上形成β'相，添加Ag元素后，会同时在基面上形成纳米级的γ''相，进一步提高合金的强度。马爱斌课题组[66−67]对Mg-10.6Gd-2Ag合金的等通道转角挤压进行了系统的研究，发现该合金在350 ℃条件下变形时可显著细化合金晶粒，同时伴随着颗粒状β(含Ag的Mg_5Gd相)的析出，时效后出现了大量γ''和β'沉淀，导致强度提高，沿ED方向的抗拉强度和屈服强度为502 MPa和409 MPa[67]。当进行8道次挤压后，合金的晶粒尺寸可细化至2 μm以下，但继续增加挤压道次对合金的晶粒细化效果并无影响，反而会引起β相的长大，导致纳米级沉淀相β'和γ''相减少，使合金强度降低[68]。挤压温度是影响合金微观组织的另一关键因素，当挤压温度降低到300 ℃时，由于温度较低，难以发生完全动态再结晶，形成了变形晶粒和再结晶细晶粒构成的混晶组织，但再结晶晶粒的尺寸较350 ℃挤压条件下更低，为0.76 μm。变形晶粒中高的位错密度使其强化，再结晶晶粒又可在变形中储存大量的位错，使合金强度和塑性协同提高，并表现出各向同性[65]。

近年来，LIU等[69−71]对Mg-Gd-Y-Zn-Zr合金ECAP变形进行了大量的研究。研究发现，Y元素含量对合金的微观组织具有重要的影响，当Y含量为2%时，铸态合金由α-Mg、Mg_3(Gd，Y，Zn)和14H-LPSO相组成；当Y含量增加到4%时，开始出现富Y相；当Y含量继续增加到6%时，合金中产生了$Mg_{24}Y_5$相、富Y相和18R-LPSO相，但未出现Mg_3(Gd，Y，Zn)。随着ECAP的进行，Y含量为2%和4%合金中的Mg_3(Gd，Y，Zn)被破碎成颗粒，不同Y元素含量合金中的14H-LPSO结构在变形过程中的演变规律则不尽相同。在Y含量为2%的合金中，14H-LPSO相仅发生弯曲，说明其塑性较好，14H-LPSO相片层内塞积了大量位错，可以有效阻止位错运动。当Y含量为4%时，14H-LPSO相先发生弯曲并最终被细化成颗粒，细化形成的粒子可通过粒子激发形核(PSN)机制促进DRX的发生[74]。当Y含量为6%时，$Mg_{24}Y_5$相破碎成颗粒，18R-LPSO部分发生弯曲和扭折，部分发生破碎后与$Mg_{24}Y_5$破碎颗粒混合在一起阻碍了位错运动，进而促进动态再结晶；在Y含量为6%合金的ECAP过程中，还析出了14H-LPSO片层。破碎的$Mg_{24}Y_5$颗粒和LPSO相因其更高的硬度和模量，对合金有增强作用。此外，变形道次还会影响合金的相转变，当4%Y的合金经过16道次挤压后，可直接析出γ'相，而在低于16道次时并未观察到该相。γ'相对合金强度的提升被认为优于LPSO相，因此，合金获得了548 MPa的抗压强度[70]。先前的研究表明，Mg-Gd-Zn合金中沉淀相的析出序列为：过饱和固溶体(Super-saturated solid solution, SSSS)→γ''($Mg_{70}Gd_{15}Zn_{15}$)→γ' (MgGdZn)→γ(Mg_{12}GdZn)”[75]。然而，在该研究中，γ'相是从基体中直接析出，γ'相的形核和长大需要热激活，虽然360℃的挤压温度高于合金通常的时效温度，但是挤压一个道次的时间太短，所以在道次较少时无法提供γ'相生成所需要的能量，而经过16道次挤压后，可累积足够的能量促使其析出。

等通道转角挤压能提供强烈的剪切，有效细化晶粒和合金中的LPSO相，获得较为优异的力学性能，但往往需要进行8道次以上的变形，导致制备工艺较复杂，成本较高，且该方法制备样品的尺寸往往有限。

3.2 高压扭转

细化晶粒是获得高强度镁合金的有效方法之一。高压扭转(High pressure torsion, HPT)技术被认为是最能细化镁合金晶粒，制备超细晶晶粒和纳米晶的最有效的大塑性变形技术[76]。近年来，研究人员通过高压扭转制备了一系列硬度超过110 HV或1100 MPa的Mg-RE合金(见表4[77−89])。

近年来，SUN等[77−83]采用高压扭转的变形方法对Mg-Gd-Y-Zn-Zr合金进行了系统的研究，研究发现，HPT对合金的晶粒具有显著的细化作用，随着HPT圈数的增加，可以将合金的晶粒细化至100 nm以下，获得超细的纳米级晶粒，样品的最高硬度可达到156 HV。

SUN等[77−78, 82]分别对铸态和固溶态Mg-8.2Gd-3.8Y-1.0Zn-0.4Zr合金进行了HPT变形，并研究了固溶态样品HPT后的时效行为。研究表明，铸态合

表4　HPT变形后硬度超过110 HV或1100 MPa的Mg-RE合金[77−89]

Table 4　Mg-RE alloy with hardness exceeding 110 HV or 1100 MPa after HPT[77−89]

Alloy	Processing condition	Hardness	Ref.
Mg-8.2Gd-3.8Y-1.0Zn-0.4Zr	HPT at ambient temperature (1 r/min)	115 HV	[77]
	(510 ℃, 12 h)+HPT at ambient temperature (1 r/min)	126 HV	[78]
	(510 ℃, 12 h, furnace cooling)+HPT at ambient temperature (1 r/min)+annealed at (400 ℃, 1 h)	128 HV	[79]
	(510 ℃, 12 h)+pre-aged at (200 ℃, 48 h)+ HPT at ambient temperature (1 r/min)	134 HV	[80]
	(510 ℃, 12 h)+HPT at ambient temperature (1 r/min)+ annealed at 573 K/0.5 h	136 HV	[81]
	(510 ℃, 12 h)+HPT at ambient temperature (1 r/min)+ aged at (120 ℃, 12 h)	145 HV	[82]
	(510 ℃, 12 h)+pre-aged at (200 ℃, 48 h)+ HPT at ambient temperature (1 r/min)+aged at 120 ℃	156 HV	[83]
Mg-4.97Sm-0.84Ca	(530 ℃, 8 h)+HPT at room temperature (1 r/min)+ aged at (125 ℃, 8 h)	145 HV	[84]
Mg-22Gd	(530 ℃, 5 h)+HPT at room temperature	146 HV	[85]
Mg-4.09Y-2.41Nd-2.14Gd-0.56Zr	(510 ℃, 12 h)+HPT at ambient temperature (1 r/min)+ aged at (120 ℃, 22 h)	138 HV	[86]
Mg-11Y	(500 ℃, 12 h)+HPT at room temperature (1.5 r/min)	135 HV	[87]
Mg-3.56Y-2.20Nd-0.47Zr	(525 ℃, 8 h)+HPT at room temperature (1 r/min)+ aged at (200 ℃, 1 h)	1411 MPa	[88]
	(525 ℃, 8 h)+HPT at room temperature (1 r/min)+aged at 200 ℃	1400 MPa	[89]

金在HPT变形后晶粒被细化至55 nm。此外，HPT处理在合金中引入了大量的位错。纳米晶粒、高密度位错和Mg_3(Gd，Y)颗粒导致合金的硬度达到了115 HV[77]。HPT前进行固溶处理可使合金获得更显著的晶粒细化效果，晶粒尺寸减小到48 nm，且增强了溶质−位错/位错−位错的相互作用，提高了位错密度，硬度达到126 HV[78]。ČÍŽEK等[85]也发现合金固溶后经HPT变形可以引入大量位错，这些位错互相缠结分布在基体中，使Mg-22Gd的合金硬度达到了145 HV。HPT引入的大量晶界和位错有利于合金元素的扩散，可加速合金的时效过程。与通常的时效工艺相比，HPT合金所需的时效温度更低，到达峰值所需的时间更短[82, 84, 86, 89]。例如，Mg-Gd合金通常的时效析出序列为“过饱和固溶体(Super-saturated solid solution, SSSS)→β''→β'→β_1→β”[75, 90]，而HPT处理的Mg-8.2Gd-3.8Y-1.0Zn-0.4Zr合金在200 ℃时效时，未观察到亚稳相生成，直接形成了平衡相β相。当在120 ℃(远低于常规时效温度：200~250 ℃)时效12 h后，即达到了145 HV的峰值硬度[82]。

第二相对镁合金的塑性变形过程具有显著的影响，SUN等[79−80]分别通过时效处理和固溶后的冷却工艺调控，在Mg-8.2Gd-3.8Y-1.0Zn-0.4Zr合金中引入了纳米级和微米级第二相。研究表明，纳米级析出相β'(Mg_7RE)有助于晶粒细化，能够在变形过程中抑制位错运动，16圈后晶粒细化至33 nm。微米级14H-LPSO相在变形过程中发生扭折，在扭折带附近累积的大量位错可为动态再结晶提供能量，晶粒细化优先发生在片状14H-LPSO附近，获得了52 nm的超细晶(见图3[79])。显然，与微米级第二相相比，纳米级第二相对晶粒的细化效果更为明显。随着变形的进行，纳米级和微米级第二相将逐步溶入基体并形成过饱和固溶体，提供固溶强化效果。SUN等[83]对固溶+时效+HPT工艺制备的样品在

120 ℃进行了再时效，使硬度进一步提升，获得了156 HV的超高硬度。

SUN等[81]采用固溶+HPT的工艺制备了晶粒尺寸为48 nm的超细晶，并研究了合金的退火行为。当在473 K、573 K退火时，可在晶界附近产生大量的溶质偏析和团簇，且晶粒尺寸基本保持不变；随退火温度继续升高，晶粒发生明显长大并析出平衡相β(Mg_5RE)，β相的析出消耗了溶质原子，使晶界处的溶质偏析明显减少，导致硬度降低。在573 K退火时合金的力学性能最为优异，硬度为136 HV，超细的晶粒、大量的溶质偏析及团簇是合金具有较高硬度的原因。

镁−稀土合金的高压扭转变形主要集中在Mg-8.2Gd-3.8Y-1.0Zn-0.4Zr合金，其他合金研究较少。通过高压扭转变形，可将晶粒尺寸显著细化至纳米级，并引入高密度位错。通过调控工艺还可引入大量第二相使合金强化，但采用该方法所制备的样品尺寸受到了极大的限制，难以用于制备大型结构件。

3.3 多向锻造

多向锻造技术(MDF)在镁合金领域应用广泛，相比其他的SPD技术，多向锻造技术的成本低且可以适用于大型工业金属产品，设备和操作方法简单。近年来，研究人员采用多向锻造变形开发出了强度超过350 MPa或是硬度超过150 Hv的Mg-RE合金(见表5[30−32, 91−97])。

MDF时，温度是影响合金的重要因素。锻造温度低，塑性变形能力差，合金容易开裂。此外，温度太低导致合金出现动态析出，消耗了稀土元素，降低了时效强化效果[98]。锻造温度升高，发生动态再结晶，使合金的变形能力提升，可防止裂纹出现。LIU等[95]对Mg-8Gd-3Y-0.4Zr合金在230~480 ℃进行自由多向锻造后发现，高于380 ℃时由

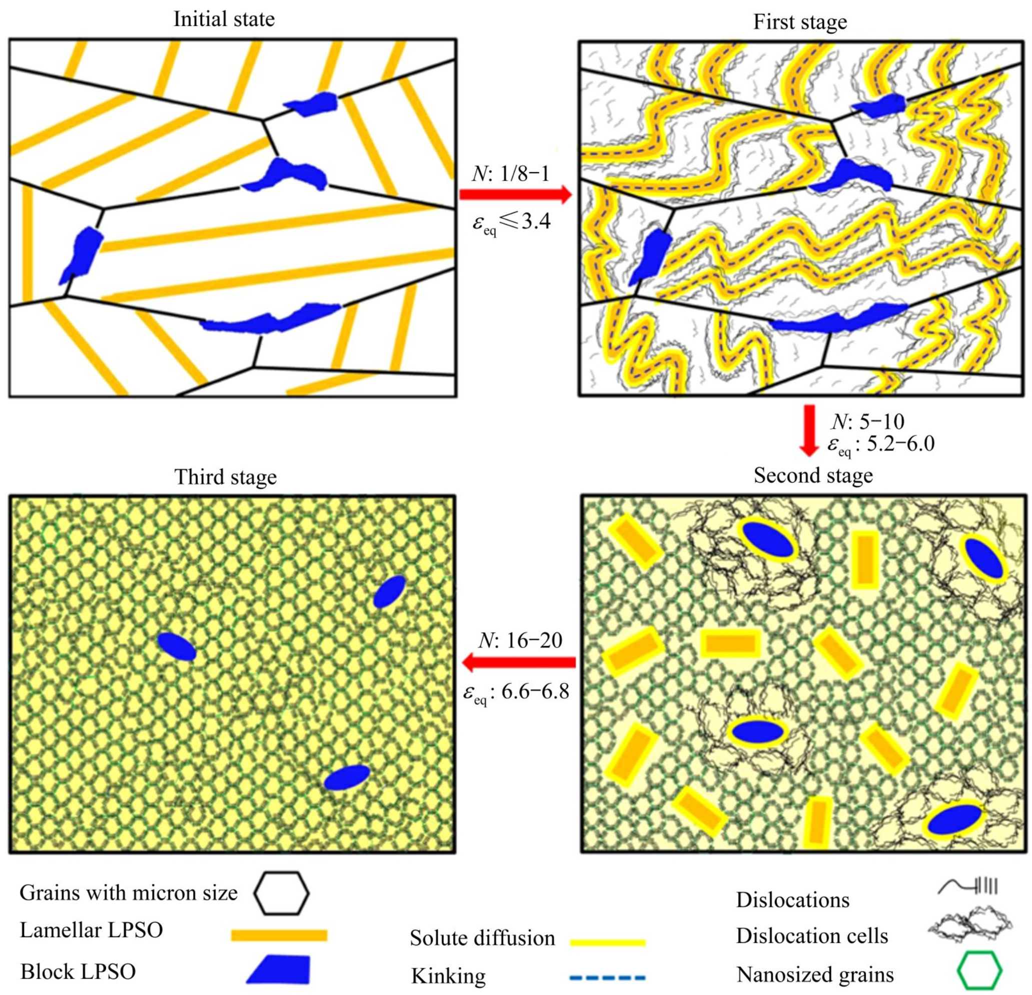

图3 微米级第二相在高压扭转过程中的演变过程示意图[79]

Fig. 3 Schematic diagram of evolution process of micron-second phase during high pressure torsion[79]

表5　MDF变形制备的Mg-RE合金的力学性能[30−32, 91−97]

Table 5　Mechanical properties of Mg-RE alloys prepared by MDF deformation[30−32, 91−97]

Alloy	Processing condition	UTS/ MPa	YS/ MPa	EL/ %	Ref.
Mg-6Gd-3Y-1.4Zn-0.4Zr	(510 ℃, 24 h)+forged at 480 ℃+ aged at (200 ℃, 26 h)	431	392	8	[32]
Mg-8.0Gd-3.7Y-0.3Ag-0.4Zr	(525 ℃, 7 h)+forged at 475 ℃+ aged at (200 ℃, 40 h)	448	391	3.9	[31]
Mg-4.9Gd-3.2Y-1.1Zn-0.5Zr	(510 ℃, 24 h)+forged at 500 ℃+ aged at (200 ℃, 78 h)	390	331	9.1	[30]
Mg-13Gd-4Y-2Zn-0.5Zr	(520 ℃, 16 h)+forged at 440 ℃	399	308	14.1	[91]
Mg-9.02Gd-4.21Y-0.48Zr	(525 ℃, 10 h)+forged at 450 ℃+Aged	365	—	—	[92]
Mg-13Gd-4Y-2Zn-0.6Zr	(520 ℃, 16 h)+forged at 480 ℃	416	342	4.12	[93]
Mg-13Gd-4Y-2Zn-0.5Zr	(520 ℃, 16 h)+forged at (480−420 ℃, temperature of each pass is reduced by 20 ℃); annealed at 480 ℃ between passes	405	305	13.1	[94]
Mg-8Gd-3Y-0.4Zr	(500 ℃, 9 h)+forged at 430 ℃	353	276	9.25	[95]
Mg-5.95Gd-3.21Y-0.83Zn-0.32Ag-0.44Zr	(480 ℃, 8 h)+(510 ℃, 24 h)+forged at 480 ℃ (9 passes)+forged at 460 ℃ (9 passes)	364	295	4.8	[96]
Mg-13Gd-4Y-2Zn-0.5Zr	(520 ℃, 16 h)+forged at 480 ℃ (1 passes)+ forged at 430 ℃ (3 passes)	Microhardness(HV) 157 HV			[97]

UTS: Ultimate tensile strength; YS: Yield strength; EL: Elongation.

于滑移系增多和动态再结晶的软化效应，明显提高了合金的可锻性。然而锻造温度过高，或是每道次间重新加热至初始锻造温度，会引起晶粒长大，导致合金强度降低[93, 99]。为此，研究者们开发了降温锻造工艺。例如，DONG等[91]对Mg-13Gd-4Y-2Zn-0.5Zr合金在480~420 ℃进行降温锻造，获得了4.0 μm的细晶粒；合金的抗拉强度、屈服强度和伸长率分别达到357 MPa、294 MPa和18.1%；时效后进一步提升到399 MPa、308 MPa和14.1%。这是由于初始阶段的高温诱发动态再结晶，随后的低温变形阶段，抑制了再结晶晶粒的长大，从而可以获得细小的晶粒组织。

由于镁合金塑性变形能力差，多向锻造变形时通常采用低的应变速率。相比于低应变速率多向锻造，高应变速率多向冲击锻造(Multi-directional impact forged, MDIF)具有更高的锻造效率，有利于锻件的大规模生产[100−101]。此外，高应变速率多向锻造对合金的晶粒细化机制和织构有显著的影响[92, 102−103]。研究发现，高应变速率下进行多向锻造变形，晶粒细化发生在两个阶段。第一阶段，刚开始变形时的应变速率太大，原子扩散和位错的运动短时间内的无法发生，孪晶被激活。且由于$\{10\bar{1}2\}$拉伸孪晶的临界分切应力低，因此，激活的几乎主要是$\{10\bar{1}2\}$拉伸孪晶。该阶段的晶粒细化主要是孪晶细化，这些孪晶将初始粗晶粒分割，晶粒尺寸显著下降，同时，初始粗晶的晶粒间也有少量的不连续动态再结晶晶粒，也有助于细化晶粒；继续变形时，由于应变的积累，位错滑移成为主要的变形机制，发生动态再结晶，因此，第二阶段的晶粒细化主要是再结晶细化机制[92, 102−105]。高应变速率多向锻造对合金的织构也有影响。先前的研究表明，由于孪晶和基面滑移的存在，镁合金在变形时会形成强基面织构[106]。然而之后的研究发现，高应变速率多向锻造后的合金主要呈随机的非基面弱织构分布。这是由于高应变速率多向锻造每道次的应变小，加上锻造方向的改变抑制了强基面织构的形成，且在随后的动态再结晶阶段继续形成更细的随机取向晶粒，因此，整体织构强度弱[92, 102−103, 105]。

此外，丁宁等[104]发现 Mg-8Gd-1Er-0.5Zr 合金高应变速率多向锻造时，由于晶粒定向形核−选择长大以及特殊受力条件下晶粒的连续旋转，导致最终形成了双峰织构。

在多向锻造道次间引入退火和预固溶工艺，可进一步优化合金的组织和性能。镁合金塑性变形时，均匀的微观组织有利于获得良好的强度。DONG 等[94]提出每道次间进行退火处理的高低温循环 MDF 变形，发现每道次后的退火处理可以通过静态再结晶(SRX)机制使合金的组织变得均匀。1 道次只发生了部分动态再结晶，在退火后组织变得均匀，可认为是等轴晶粒；2 道次后，析出了颗粒和板状的 Mg_5(Gd,Y,Zn)，退火后大部分 Mg_5(Gd,Y,Zn)被溶解；随后继续变形，最终获得了均匀的组织，晶粒细化至 5.2 μm。DONG 等[97]在每道次前对含 LPSO 相的合金进行预固溶处理，发现在预固溶以及之后的多向锻造变形过程中，晶粒内部析出了大量对合金起强化作用的片状 Mg_5RE 相(见图 4[97])。此外，经预固溶处理和多向锻造变形，块状 LPSO 相也向层状 LPSO 相转变。两者共同作为强化相，使合金的硬度提高至 157 HV。

综上所述，降温多向锻造能显著抑制再结晶晶粒长大，高应变速率多向锻造可以改变晶粒细化机制和弱化织构，锻造道次间引入退火和预固溶工艺可以使显微组织均匀化并析出强化相，可提高合金的综合性能，但多向锻造的工艺较复杂，生产成本较高。

3 结论与展望

本文介绍了常规挤压、常规轧制以及大塑性变形技术对镁−稀土合金微观组织和力学性能的影响，发现提高合金力学性能的方法主要包括细化晶粒、形成双峰分布晶粒和引入层错三个方面。从现有的研究来看，对晶粒细化方面的研究更为广泛，而涉及形成双峰分布晶粒和引入层错等方面的研究还相对较少。在未来，研究范围应该进一步拓展和深化，对此，本文对高性能镁−稀土合金塑性变形技术的研究提出如下建议。

1) 在常规挤压方面，随着研究的深入，通过合金成分和变形工艺的调整使晶粒细化可能会出现瓶

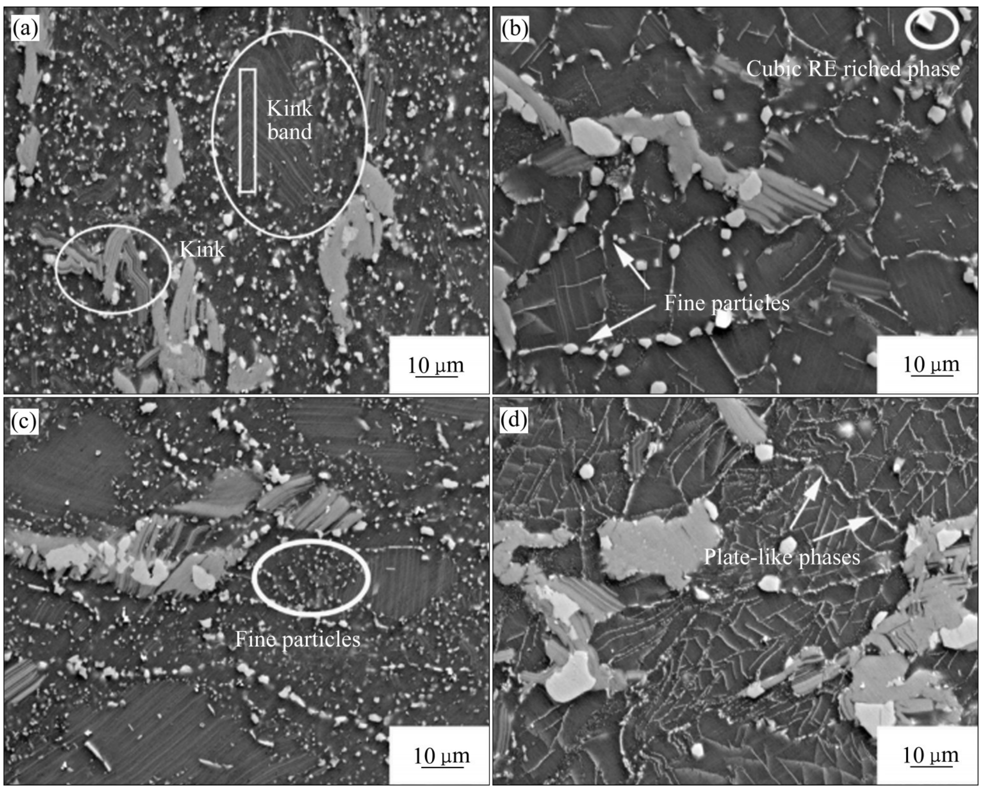

图4 不同道次后的 MDF 样品的 SEM 像[97]

Fig. 4 SEM images of MDF samples after different passes[97]: (a) 1 passes; (b) 2 passes; (c) 3 passes; (d) 4 passes

颈，今后可在新的强化机制如形成双峰分布晶粒等方面开展更多、更深入的研究。例如，深入研究初始组织、变形工艺和新的模具设计等对双峰分布晶粒形成规律的影响以及双峰分布晶粒影响合金强韧性的机理。

2) 在常规轧制方面，应该以形成双峰分布晶粒和层错为目标，探索合金成分、初始组织、变形工艺和热处理工艺等对两种组织的形成规律，并深入开展层错影响合金力学性能机理方面的研究。

3) 在大塑形变形技术方面，虽然晶粒细化效果明显，但所获得的强化效果往往不够理想，需要深入研究晶粒细化与其他多种强化机制的协同。在制备的样品尺寸上，仍然需要进一步突破大尺寸样品的制备技术，复杂的变形工艺也需要进一步简化。

REFERENCES

[1] 付彭怀, 彭立明, 丁文江. 汽车轻量化技术: 铝/镁合金及其成型技术发展动态[J]. 中国工程科学, 2018, 20(1): 84−90.
FU Peng-huai, PENG Li-ming, DING Wen-jiang. Automobile lightweight technology: Development trends of aluminum/magnesium alloys and their forming technologies[J]. Strategic Study of CAE, 2018, 20(1): 84−90.

[2] 刘婷婷, 潘复生. 镁合金“固溶强化增塑”理论的发展和应用[J]. 中国有色金属学报, 2019, 29(9): 2050−2063.
LIU Ting-ting, PAN Fu-sheng. Development and application of “solid solution strengthening and ductilizing” for magnesium alloys[J]. The Chinese Journal of Nonferrous Metals, 2019, 29(9): 2050−2063.

[3] ZENG Z R, STANFORD N, DAVIES C H J, et al. Magnesium extrusion alloys: A review of developments and prospects[J]. International Materials Reviews, 2019, 64(1): 27−62.

[4] XU T C, YANG Y, PENG X D, et al. Overview of advancement and development trend on magnesium alloy[J]. Journal of Magnesium and Alloys, 2019, 7(3): 536−544.

[5] KONG L. Two main and a new type rare earth elements in Mg alloys: A review[C]. IOP Conference Series: Materials Science and Engineering. Changsha: IOP Publishing, 2017: 012026.

[6] 王敬丰, 彭 星, 王 奎, 等. 超大规格宽幅薄壁中空镁合金型材挤压成形的数值模拟及实验研究[J]. 中国有色金属学报, 2020, 30(12): 2809−2819.
WANG Jing-feng, PENG Xing, WANG Kui, et al. Numerical simulation and experimental study on extrusion forming of ultra-large size wide thin-walled hollow magnesium alloy profiles[J]. The Chinese Journal of Nonferrous Metals, 2020, 30(12): 2809−2819.

[7] 丁文江, 吴玉娟, 彭立明, 等. 高性能镁合金研究及应用的新进展[J]. 中国材料进展, 2010, 29(8): 37−45, 36.
DING Wen-jiang, WU Yu-juan, PENG Li-ming, et al. Research and application development of advanced magnesium alloys[J]. Materials China, 2010, 29(8): 37−45, 36.

[8] ROKHLIN L L. Magnesium alloys containing rare earth metals: Structure and properties[M]. London: CRC Press, 2003.

[9] HE S M, ZENG X Q, PENG L M, et al. Microstructure and strengthening mechanism of high strength Mg-10Gd-2Y-0.5Zr alloy[J]. Journal of Alloys and Compounds, 2007, 427(1/2): 316−323.

[10] ZHANG J H, LIU S J, WU R Z, et al. Recent developments in high-strength Mg-RE-based alloys: Focusing on Mg-Gd and Mg-Y systems[J]. Journal of Magnesium and Alloys, 2018, 6(3): 277−291.

[11] 李 芳, 管仁国, 铁 镝, 等. 我国先进镁合金材料产业2035发展战略研究[J]. 中国工程科学, 2020, 22(5): 76−83.
LI Fang, GUAN Ren-guo, TIE Di, et al. Development strategies for China’s advanced magnesium alloy industry toward 2035[J]. Strategic Study of CAE, 2020, 22(5): 76−83.

[12] 候正全, 蒋 斌, 王煜烨, 等. 镁合金新材料及制备加工新技术发展与应用[J]. 上海航天(中英文), 2021, 38(3): 119−133.
HOU Zheng-quan, JIANG Bin, WANG Yu-ye, et al. Development and application of new magnesium alloy materials and their new preparation and processing technologies[J]. Aerospace Shanghai (Chinese & English), 2021, 38(3): 119−133.

[13] 丁文江, 吴国华, 李中权, 等. 轻质高性能镁合金开发及其在航天航空领域的应用[J]. 上海航天(中英文), 2019, 36(2): 1−8.
DING Wen-jiang, WU Guo-hua, LI Zhong-quan, et al. Development of high-performance light-mass magnesium alloys and applications in aerospace and aviation fields[J]. Aerospace Shanghai (Chinese & English), 2019, 36(2): 1−8.

[14] WAN Y C, TANG B, GAO Y H, et al. Bulk nanocrystalline high-strength magnesium alloys prepared via rotary swaging[J]. Acta Materialia, 2020, 200: 274−286.

[15] YOU S H, HUANG Y D, KAINER K U, et al. Recent research and developments on wrought magnesium alloys[J]. Journal of Magnesium and Alloys, 2017, 5(3): 239−253.

[16] 宋 波, 辛仁龙, 郭 宁, 等. 变形镁合金室温应变硬化行为的研究进展[J]. 中国有色金属学报, 2014, 24(11): 2699−2710.
SONG Bo, XIN Ren-long, GUO Ning, et al. Research progress of strain hardening behavior at room temperature in

wrought magnesium alloys[J]. The Chinese Journal of Nonferrous Metals, 2014, 24(11): 2699−2710.

[17] 丁文江, 靳 丽, 吴文祥, 等. 变形镁合金中的织构及其优化设计[J]. 中国有色金属学报, 2011, 21(10): 2371−2381.
DING Wen-jiang, JIN Li, WU Wen-xiang, et al. Texture and texture optimization of wrought Mg alloy[J]. The Chinese Journal of Nonferrous Metals, 2011, 21(10): 2371−2381.

[18] YU H H, XIN Y C, WANG M Y, et al. Hall-Petch relationship in Mg alloys: A review[J]. Journal of Materials Science & Technology, 2018, 34(2): 248−256.

[19] SHI B Q, CHENG Y Q, SHANG X L, et al. Hall-Petch relationship, twinning responses and their dependences on grain size in the rolled Mg-Zn and Mg-Y alloys[J]. Materials Science and Engineering A, 2019, 743: 558−566.

[20] JIN Z Z, ZHA M, YU Z Y, et al. Exploring the Hall-Petch relation and strengthening mechanism of bimodal-grained Mg-Al-Zn alloys[J]. Journal of Alloys and Compounds, 2020, 833: 155004.

[21] TANG C P, LIU W H, CHEN Y Q, et al. Effects of thermal treatment on microstructure and mechanical properties of a Mg-Gd-based alloy plate[J]. Materials Science and Engineering A, 2016, 659: 63−75.

[22] 唐昌平, 李国栋, 刘文辉, 等. 析出相对 Mg-Gd-Y-Nd-Zr 合金室温压缩行为的影响[J]. 材料导报, 2017, 31(16): 103−106.
TANG Chang-ping, LI Guo-dong, LIU Wen-hui, et al. Effects of precipitates on compression behavior of Mg-Gd-Y-Nd-Zr alloy at ambient temperature[J]. Materials Reports, 2017, 31(16): 103−106.

[23] WANG K, DOU X X, WANG J F, et al. Achieving enhanced mechanical properties in Mg-Gd-Y-Zn-Mn alloy by altering dynamic recrystallization behavior via pre-ageing treatment[J]. Materials Science and Engineering A, 2020, 790: 139635.

[24] YU Z J, XU C, MENG J, et al. Microstructure evolution and mechanical properties of as-extruded Mg-Gd-Y-Zr alloy with Zn and Nd additions[J]. Materials Science and Engineering A, 2018, 713: 234−243.

[25] WANG N, YANG Q, LI X L, et al. Microstructures and mechanical properties of a Mg-9Gd-3Y-0.6Zn-0.4Zr (wt.%) alloy modified by Y-rich misch metal[J]. Materials Science and Engineering A, 2021, 806: 140609.

[26] SU N, XUE X Y, ZHOU H, et al. Effects of nanoprecipitates and LPSO structure on deformation and fracture behaviour of high-strength Mg-Gd-Y-Zn-Mn alloys[J]. Materials Characterization, 2020, 165: 110396.

[27] LIU S J, WANG K, WANG J F, et al. Ageing behavior and mechanisms of strengthening and toughening of ultrahigh-strength Mg-Gd-Y-Zn-Mn alloy[J]. Materials Science and Engineering A, 2019, 758: 96−98.

[28] GUAN K, YANG Q, BU F Q, et al. Microstructures and mechanical properties of a high-strength Mg-3.5Sm-0.6Zn-0.5Zr alloy[J]. Materials Science and Engineering A, 2017, 703: 97−107.

[29] YANG Z Q, MA A, XU B Q, et al. Development of a high-strength Mg-11Gd-2Ag (wt%) alloy sheet with extra-low anisotropy[J]. Materials Science and Engineering A, 2021, 811: 141084.

[30] ZHOU X J, YAO Y, ZHANG J, et al. A high-performance Mg-4.9Gd-3.2Y-1.1Zn-0.5Zr alloy via multidirectional forging after analyzing its compression behavior[J]. Journal of Materials Science & Technology, 2021, 70: 156−167.

[31] WANG B Z, LIU C M, GAO Y H, et al. Microstructure evolution and mechanical properties of Mg-Gd-Y-Ag-Zr alloy fabricated by multidirectional forging and ageing treatment[J]. Materials Science and Engineering A, 2017, 702: 22−28.

[32] HUANG C, LIU C M, WANG B Z. Microstructures and Tensile Properties of Mg-6Gd-3Y-1.4Zn-0.4Zr Alloys[J]. Materials Science Forum, 2016, 849: 148−153.

[33] LI B S, GUAN K, YANG Q, et al. Effects of 0.5wt% Ce addition on microstructures and mechanical properties of a wrought Mg-8Gd-1.2Zn-0.5Zr alloy[J]. Journal of Alloys and Compounds, 2018, 763: 120−133.

[34] ZHANG D D, YANG Q, GUAN K, et al. A high-strength low-rare-earth-alloyed magnesium alloy via traditional hot-extrusion[J]. Journal of Alloys and Compounds, 2019, 810: 151967.

[35] RONG W, ZHANG Y, WU Y J, et al. Fabrication of high-strength Mg-Gd-Zn-Zr alloys via differential-thermal extrusion[J]. Materials Characterization, 2017, 131: 380−387.

[36] YANG Z, XU C, NAKATA T, et al. Effect of extrusion ratio and temperature on microstructures and tensile properties of extruded Mg-Gd-Y-Mn-Sc alloy[J]. Materials Science and Engineering A, 2021, 800: 140330.

[37] YU Z J, XU C, MENG J, et al. Effects of extrusion ratio and temperature on the mechanical properties and microstructure of as-extruded Mg-Gd-Y- (Nd/Zn) -Zr alloys[J]. Materials Science and Engineering A, 2019, 762: 138080.

[38] HUANG H, MIAO H W, YUAN G Y, et al. Fabrication of ultra-high strength magnesium alloys over 540 MPa with low alloying concentration by double continuously extrusion[J]. Journal of Magnesium and Alloys, 2018, 6(2): 107−113.

[39] LIU X, ZHANG Z Q, HU W Y, et al. Effects of extrusion speed on the microstructure and mechanical properties of Mg-9Gd-3Y-1.5Zn-0.8Zr alloy[J]. Journal of Materials Science & Technology, 2016, 32(4): 313−319.

[40] LIU W, ZENG Z R, HOU H, et al. Dynamic precipitation behavior and mechanical properties of hot-extruded $Mg_{89}Y_4Zn_2Li_5$ alloys with different extrusion ratio and speed[J]. Materials Science and Engineering A, 2020, 798: 140121.

[41] MA H, HUANG Z H, YAO Y, et al. Evolution of microstructures and mechanical properties of Mg-1.4Gd-1.2Y-0.4Zn-0.5Al sheets with different extrusion ratios[J]. Journal of Alloys and Compounds, 2020, 817: 152769.

[42] XU C, NAKATA T, QIAO X G, et al. Effect of extrusion parameters on microstructure and mechanical properties of Mg-7.5Gd-2.5Y-3.5Zn-0.9Ca-0.4Zr (wt%) alloy[J]. Materials Science and Engineering A, 2017, 685: 159−167.

[43] LI R G, LI H R, PAN H C, et al. Achieving exceptionally high strength in binary Mg-13Gd alloy by strong texture and substantial precipitates[J]. Scripta Materialia, 2021, 193: 142−146.

[44] LI R G, NIE J F, HUANG G J, et al. Development of high-strength magnesium alloys via combined processes of extrusion, rolling and ageing[J]. Scripta Materialia, 2011, 64(10): 950−953.

[45] ZHAO T S, HU Y B, ZHANG C, et al. Influence of extrusion conditions on microstructure and mechanical properties of Mg-2Gd-0.3Zr magnesium alloy[J]. Journal of Magnesium and Alloys, 2020, 10(2): 387−399.

[46] 任聪林, 宽　军, 曹　鑫, 等. 挤压速度对Mg-2Y-1Zn-0.4Zr合金组织演变、力学性能和动态腐蚀行为的影响[J]. 材料热处理学报, 2020, 41(9): 117−125.
REN Cong-lin, KUAN Jun, CAO Xin, et al. Effect of extrusion speed on microstructure evolution, mechanical properties and dynamic corrosion behavior of Mg-2Y-1Zn-0.4Zr alloy[J]. Transactions of Materials and Heat Treatment, 2020, 41(9): 117-125.

[47] 万　佳, 刘楚明, 许诗源, 等. 挤压比对Mg-Gd-Ni合金微观组织、力学及腐蚀性能的影响[J]. 中南大学学报自然科学版, 2020, 51(9): 2405−2412.
WAN Jia, LIU Chu-ming, XU Shi-yuan, et al. Effect of extrusion ratio on microstructure, mechanical and corrosion properties of Mg-Gd-Ni alloy[J]. Journal of Central South University (Science and Technology) , 2020, 51(9): 2405−2412.

[48] JIAN W W, CHENG G M, XU W Z, et al. Ultrastrong Mg alloy via nano-spaced stacking faults[J]. Materials Research Letters, 2013, 1(2): 61−66.

[49] LI B, TENG B G, WANG E D. Effects of accumulative rolling reduction on the microstructure characteristic and mechanical properties of Mg-Gd-Y-Zn-Zr sheets processed by hot rolling[J]. Materials Science and Engineering A, 2019, 765: 138317.

[50] 刘栩东, 吴皓月, 刘秀兰, 等. 热轧及退火处理对Mg-10Gd-3Y-0.3Zr合金显微组织及抗压强度的影响[J]. 西安工业大学学报, 2018, 38(1): 58−63.
LIU Xu-dong, WU Hao-yue, LIU Xiu-lan, et al. Influence of hot rolling and annealing treatment on microstructure and compressive stress of Mg-10Gd-3Y-0.3Zr alloy[J]. Journal of Xi'an Technological University, 2018, 38(1): 58−63.

[51] SOMEKAWA H, YAMASAKI M, KAWAMURA Y, et al. Wrought-procedure memory in caliber rolled Mg-Y-Zn alloy containing LPSO phase[J]. Materials Characterization, 2021, 175: 111080.

[52] XU C, ZHENG M Y, XU S W, et al. Ultra high-strength Mg-Gd-Y-Zn-Zr alloy sheets processed by large-strain hot rolling and ageing[J]. Materials Science and Engineering A, 2012, 547: 93−98.

[53] SUN J P, LI B J, YUAN J, et al. Developing a high-performance Mg-5.7Gd-1.9Ag wrought alloy via hot rolling and aging[J]. Materials Science and Engineering A, 2021, 803: 140707.

[54] ULLMANN M, KITTNER K, HENSELER T, et al. Dynamic recrystallization and texture evolution of Mg-6.8Y-2.5Zn-0.3Zr alloy during hot rolling[J]. Procedia Manufacturing, 2020, 50: 809−816.

[55] WANG J H, JIN P P, LI X Q, et al. Effect of rolling with different amounts of deformation on microstructure and mechanical properties of the Mg-1Al-4Y alloy[J]. Materials Characterization, 2020, 161: 110149.

[56] XIN Y, WANG M Y, ZENG Z, et al. Strengthening and toughening of magnesium alloy by $\{10\bar{1}2\}$ extension twins [J]. Scripta Materialia, 2012, 66(1): 25−28.

[57] NEH K, ULLMANN M, KAWALLA R. Mechanical properties and microstructure of the magnesium alloy Mg-6.8Y-2.5Zn-0.5Al produced by casting and hot rolling[J]. Materials Science Forum, 2018, 918: 3−12.

[58] CABIBBO M, PAOLETTI C, MINÁRIK P, et al. Secondary phase precipitation and thermally stable microstructure refinement induced by ECAP on Mg-Y-Nd (WN43) alloy[J]. Materials Letters, 2019, 237: 5−8.

[59] ZHANG Z, ZHANG J H, WANG J, et al. Toward the development of Mg alloys with simultaneously improved strength and ductility by refining grain size via the deformation process[J]. International Journal of Minerals, Metallurgy and Materials, 2021, 28(1): 30−45.

[60] SIVASHANMUGAM N, HARIKRISHNA K L. Influence of rare earth elements in magnesium alloy—A mini review[J]. Materials Science Forum, 2020, 979: 162−166.

[61] LIU H, JU J, LU F M, et al. Dynamic precipitation behavior and mechanical property of an $Mg_{94}Y_4Zn_2$ alloy prepared by multi-pass successive equal channel angular pressing[J].

Materials Science and Engineering A, 2017, 682: 255−259.

[62] LIU H, JU J, YANG X W, et al. A two-step dynamic recrystallization induced by LPSO phases and its impact on mechanical property of severe plastic deformation processed $Mg_{97}Y_2Zn_1$ alloy[J]. Journal of Alloys and Compounds, 2017, 704: 509−517.

[63] LIU H, CHENG Z J, YAN K, et al. Effect of multi-pass equal channel angular pressing on the microstructure and mechanical properties of a heterogeneous $Mg_{88}Y_8Zn_4$ alloy[J]. Journal of Materials Science & Technology, 2016, 32(12): 1274−1281.

[64] ZHANG J S, CHEN C J, CHENG W L, et al. High-strength $Mg_{93.96}Zn_2Y_4Sr_{0.04}$ alloy with long-period stacking ordered structure[J]. Materials Science and Engineering A, 2013, 559: 416−420.

[65] XU B Q, SUN J P, YANG Z Q, et al. A near-isotropic ultrafine-grained Mg-Gd-Ag alloy with high strength-ductility synergy[J]. Journal of Materials Research and Technology, 2020, 9(6): 13616−13624.

[66] SUN J P, XU B Q, YANG Z Q, et al. Achieving excellent ductility in high-strength Mg-10.6Gd-2 Ag alloy via equal channel angular pressing[J]. Journal of Alloys and Compounds, 2020, 817: 152688.

[67] YANG Z Q, MA A B, XU B Q, et al. Revealing the tensile anisotropy, tension-compression asymmetry, and strain-hardening behavior of a high-performance Mg-Gd-Ag alloy[J]. Journal of Alloys and Compounds, 2021, 868: 159238.

[68] FU Y T, SUN J P, YANG Z Q, et al. Aging behavior of a fine-grained Mg-10.6Gd-2Ag alloy processed by ECAP[J]. Materials Characterization, 2020, 165: 110398.

[69] LIU H, JU J, YANG X W, et al. Microstructure and mechanical property of Mg-10Gd-2Y-1.5Zn-0.5Zr alloy processed by eight-pass equal-channel angular pressing [EB/OL]. Rare Metals, 2018. https://doi.org/10.1007/s12598-018-1022-1.

[70] LIU H, JU J, BAI J, et al. Preparation, microstructure evolutions, and mechanical property of an ultra-fine grained Mg-10Gd-4Y-1.5Zn-0.5Zr alloy[J]. Metals, 2017, 7(10): 398.

[71] LIU H, HUANG H, YANG X W, et al. Microstructure and mechanical property of a high-strength Mg-10Gd-6Y-1.5Zn-0.5Zr alloy prepared by multi-pass equal channel angular pressing[J]. Journal of Magnesium and Alloys, 2017, 5(2): 231−237.

[72] WANG L S, JIANG J H, SALEH B, et al. Controlling corrosion resistance of a biodegradable Mg-Y-Zn alloy with LPSO phases via multi-pass ECAP process[J]. Acta Metallurgica Sinica (English Letters), 2020, 33(9): 1180−1190.

[73] WU H R, JIANG J H, LIU H, et al. A novel method for improving the strength and ductility of Mg-Y-Er-Zn alloy using rotary-die equal-channel angular pressing[J]. Journal of Materials Research and Technology, 2021, 13: 1752−1758.

[74] LI B, TENG B G, CHEN G X. Microstructure evolution and mechanical properties of Mg-Gd-Y-Zn-Zr alloy during equal channel angular pressing[J]. Materials Science and Engineering A, 2019, 744: 396−405.

[75] NIE J F. Precipitation and hardening in magnesium alloys[J]. Metallurgical and Materials Transactions A, 2012, 43(11): 3891−3939.

[76] TANG L L, ZHAO Y H, ISLAMGALIEV R K, et al. Microstructure and thermal stability of nanocrystalline Mg-Gd-Y-Zr alloy processed by high pressure torsion[J]. Journal of Alloys and Compounds, 2017, 721: 577−585.

[77] SUN W T, XU C, QIAO X G, et al. Evolution of microstructure and mechanical properties of an as-cast Mg-8.2Gd-3.8Y-1.0Zn-0.4Zr alloy processed by high pressure torsion[J]. Materials Science and Engineering A, 2017, 700: 312−320.

[78] SUN W T, QIAO X G, ZHENG M Y, et al. Microstructure and mechanical properties of a nanostructured Mg-8.2Gd-3.8Y-1.0Zn-0.4Zr supersaturated solid solution prepared by high pressure torsion[J]. Materials & Design, 2017, 135: 366−376.

[79] SUN W T, QIAO X G, ZHENG M Y, et al. Evolution of long-period stacking ordered structure and hardness of Mg-8.2Gd-3.8Y-1.0Zn-0.4Zr alloy during processing by high pressure torsion[J]. Materials Science and Engineering A, 2018, 738: 238−252.

[80] SUN W T, QIAO X G, ZHENG M Y, et al. Exceptional grain refinement in a Mg alloy during high pressure torsion due to rare earth containing nanosized precipitates[J]. Materials Science and Engineering A, 2018, 728: 115−123.

[81] SUN W T, QIAO X G, ZHENG M Y, et al. Exceptional thermal stability and enhanced hardness in a nanostructured Mg-Gd-Y-Zn-Zr alloy processed by high pressure[EB/OL]. SSRN Electronic Journal. https://doi. org/10.21391. ssrn. 3298920.

[82] SUN W T, QIAO X G, ZHENG M Y, et al. Altered ageing behaviour of a nanostructured Mg-8.2Gd-3.8Y-1.0Zn-0.4Zr alloy processed by high pressure torsion[J]. Acta Materialia, 2018, 151: 260−270.

[83] SUN W T, QIAO X G, ZHENG M Y, et al. Achieving ultra-high hardness of nanostructured Mg-8.2Gd-3.2Y-1.0Zn-0.4Zr alloy produced by a combination of high pressure torsion and ageing treatment[J]. Scripta Materialia, 2018, 155: 21−25.

[84] LIU X H, LI Y S, MAN Y, et al. Precipitation and

recrystallization of HPT-processed Mg-Sm-Ca alloy at low temperatures[J]. Materials Letters, 2020, 277: 128252.

[85] ČÍŽEK J, HRUŠKA P, VLASÁK T, et al. Microstructure development of ultra fine grained Mg-22wt% Gd alloy prepared by high pressure torsion[J]. Materials Science and Engineering A, 2017, 704: 181−191.

[86] LI Y S, QU C, WANG J H, et al. Exceptional aging hardening behaviour of nanocrystalline Mg-Y-Nd-Gd-Zr alloy prepared by high pressure torsion[J]. Journal of Alloys and Compounds, 2020, 813: 152123.

[87] CHEN X, XIAO L, YI L, et al. High strength-ductility of heterogeneous sandwich Mg-Y alloys produced by high pressure torsion[J]. Vacuum, 2020, 179: 109568.

[88] LUKYANOVA E, MARTYNENKO N, LI E V, et al., Effect of high pressure torsion on the structure, microhardness and heating behaviour of the magnesium alloy WE43[J]. Materials Science Non-Equilibrium Phase Transformation, 2017, 3(4): 161−164.

[89] LUK'YANOVA A, MARTYNENKO N, SHAKHOVA I, et al. Strengthening of age-hardenable WE43 magnesium alloy processed by high pressure torsion[J]. Materials Letters, 2016, 170: 5−9.

[90] ZHANG X M, TANG C P, DENG Y L, et al. Phase transformation in Mg-8Gd-4Y-Nd-Zr alloy[J]. Journal of Alloys and Compounds, 2011, 509(21): 6170−6174.

[91] DONG B B, ZHANG Z M, YU J M, et al. Microstructure, texture evolution and mechanical properties of multi-directional forged Mg-13Gd-4Y-2Zn-0.5Zr alloy under decreasing temperature[J]. Journal of Alloys and Compounds, 2020, 823: 153776.

[92] SHAH S S A, WU D, WANG W H, et al. Microstructural evolution and mechanical properties of a Mg-Gd-Y alloy processed by impact forging[J]. Materials Science and Engineering A, 2017, 702: 153−160.

[93] LI B, TENG B G, LUO D G. Effects of passes on microstructure evolution and mechanical properties of Mg-Gd-Y-Zn-Zr alloy during multidirectional forging[J]. Acta Metallurgica Sinica (English Letters), 2018, 31(10): 1009−1018.

[94] DONG B B, ZHANG Z M, CHE X, et al. Microstructure, texture evolution, and mechanical properties of MDFed GWZ alloy containing LPSO phases on the condition of high and low temperature cycle deformation[J]. Metals, 2020, 10(1): 136.

[95] LIU B, ZHANG Z Y, JIN L, et al. Forgeability, microstructure and mechanical properties of a free-forged Mg-8Gd-3Y-0.4Zr alloy[J]. Materials Science and Engineering A, 2016, 650: 233−239.

[96] HUANG C, LIU C M, JIANG S N, et al. Inhomogeneous microstructure and mechanical anisotropy of multi-directional forged Mg-Gd-Y-Zn-Ag-Zr alloy[J]. Materials Science and Engineering A, 2021, 807: 140853.

[97] DONG B B, CHE X, ZHANG Z M, et al. Microstructure evolution and microhardness of Mg-13Gd-4Y-2Zn-0.5Zr alloy via pre-solution and multi-directional forging (MDF) process[J]. Journal of Alloys and Compounds, 2021, 853: 157066.

[98] XIA X S, CHEN Q, ZHAO Z D, et al. Microstructure, texture and mechanical properties of coarse-grained Mg-Gd-Y-Nd-Zr alloy processed by multidirectional forging[J]. Journal of Alloys and Compounds, 2015, 623: 62−68.

[99] 丁　宁, 杜文博, 付金龙, 等. Mg-8Gd-1Er-0.5Zr 合金多向锻造工艺及锻后组织与力学性能研究[J]. 锻压技术, 2020, 45(5): 1−5.
DING Ning, DU Wen-bo, FU Jin-long, et al. Research on multi-directional forging process and microstructure and mechanical properties after forging for Mg-8Gd-1Er-0.5Zr alloy[J]. Forging & Stamping Technology, 2020, 45(5): 1−5.

[100] WU Y Z, YAN H G, CHEN J H, et al. Microstructure and mechanical properties of ZK60 magnesium alloy fabricated by high strain rate multiple forging[J]. Materials Science and Technology, 2013, 29(1): 54−59.

[101] LI J L, WU D, YANG Q B, et al. Superplasticity of multi-directional impact forged Mg-Gd-Y-Zr alloy[J]. Journal of Alloys and Compounds, 2016, 672: 27−35.

[102] SHAH S S A, JIANG M G, WU D, et al. Dynamic recrystallization and texture evolution of GW94 Mg alloy during multi- and unidirectional impact forging[J]. Acta Metallurgica Sinica (English Letters), 2018, 31(9): 923−932.

[103] SHAH S S A, WU D, CHEN R S, et al. Temperature effects on the microstructures of Mg-Gd-Y alloy processed by multi-direction impact forging[J]. Acta Metallurgica Sinica (English Letters), 2020, 33(2): 243−251.

[104] 丁　宁, 王云峰, 刘　轲, 等. 高应变速率多向锻造 Mg-8Gd-1Er-0.5Zr 合金的微观组织、织构及力学性能[J]. 金属学报, 2021, 57(8): 1000−1008.
DING Ning, WANG Yun-feng, LIU Ke, et al. Microstructure, texture, and mechanical properties of Mg-8Gd-1Er-0.5Zr alloy by multi-directional forging at high strain rate[J]. Acta Metallurgica Sinica, 2021, 57(8): 1000−1008.

[105] 王　锐, 王晓轩, 张　娜, 等. 高应变率多向锻造及热处理对 GW93 镁合金显微组织和力学性能的影响[J]. 轻合金加工技术, 2017, 45(4): 25−31.
WANG Rui, WANG Xiao-xuan, ZHANG Na, et al. Effect of high strain-rate multi-directional forging and heat treatment process on microstructure and mechanical properties of GW93 magnesium alloy[J]. Light Alloy Fabrication

Technology, 2017, 45(4): 25−31.
[106] BASU I, AL-SAMMAN T. Twin recrystallization mechanisms in magnesium-rare earth alloys[J]. Acta Materialia, 2015, 96: 111−132.

Research progress of plastic deformation technology of magnesium-rare earth alloy

CUI Lei[1], TANG Chang-ping[1, 2], LI Quan[3], LIU Xiao[1]

(1. High-efficiency Light Alloy Component Forming Technology and Damage Resistance Evaluation Hunan Engineering Research Center, Hunan Provincial Key Laboratory of Advanced Materials for New Energy Storage and Conversion, School of Materials Science and Engineering, Hunan University of Science and Technology, Xiangtan 411201, China;
2. Guangdong Key Laboratory of Precision Equipment and Manufacturing Technique, South China University of Technology, Guangzhou 510641, China;
3. Chongqing Academy of Science and Technology, Chongqing 401123, China)

Abstract: This paper reviewed the research progress of magnesium-rare earth alloys in extrusion, rolling and severe plastic deformation (SPD) technologies. The effects of deformation temperature, deformation speed and extrusion ratio on microstructure and mechanical properties of the extruded samples were introduced. It is indicated that grain refinement or formation of bimodal-grained structure through deformation process control can improve the mechanical properties of the alloy. The research progress on rolling deformation of magnesium-rare earth alloy was summarized, it is found that the ultra-high strength magnesium alloy can be fabricated by forming bimodal-grained structure or introducing stacking faults through process control. The effects of equal channel angular pressing (ECAP), high-pressure torsion (HPT) and multi-directional forging (MDF) on the microstructure and mechanical properties of alloys were summarized. It is exhibited that SPD technology, especially HPT, is an effective method for preparing nanometer-scale ultrafine grains. Compared with extrusion and rolling, the process is more complicated and expensive, and the size of the prepared samples is often smaller. Finally, the suggestions were provided for the development direction of magnesium-rare earth alloy plastic deformation technology.

Key words: Mg-RE alloys; extrusion; rolling; severe plastic deformation; mechanical property

Foundation item: Projects(52075167, 51605159, 52071139) supported by the National Natural Science Foundation of China; Project(2020JJ4307) supported by the Natural Science Foundation of Hunan Province, China; Project(19B214) supported by Excellent Youth Project of Education Department of Hunan Province, China; Project(PEMT202103) supported by Research Fund of Key Laboratory of Precision Equipment and Manufacturing Technique of Guangdong Province, China; Project(CX20211030) supported by Postgraduate Scientific Research Innovation Project of Hunan Province, China

Received date: 2021-11-04; **Accepted date:** 2022-01-18

Corresponding author: TANG Chang-ping; Tel: +86-18773260825; E-mail: tcpswnu@163.com

(编辑　李艳红)