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镁合金“固溶强化增塑”理论的发展和应用

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摘 要：镁是密排六方结构金属，滑移系较少，其基面滑移阻力比柱面和锥面滑移阻力低很多，基面滑移启动后其他滑移系很难启动，导致镁合金室温和低温的塑性变形能力较差。重庆大学等单位研究发现某些特定原子固溶在镁中既能阻碍基面位错滑移提高强度，又能通过缩小基面与非基面滑移阻力差距促进非基面滑移开启而改善塑性，达到同时提高镁合金强度和塑性的目的。重庆大学镁合金科研团队把这一结果发展为“镁合金固溶强化增塑”合金设计理论(Solid solution strengthening and ductilizing, SSSD)。这一合金设计理论在过去十几年中已成为解决镁合金强度和塑性平衡优化的一条新途径。重庆大学应用该理论开发了多种新型高性能镁合金，其中 10 多个新合金已批准为国家标准牌号合金和国际标准牌号合金。

关键词：镁合金；强度；塑性；固溶强化增塑
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随着全球资源能源危机的加剧和环境负担的增加，世界各国制定了越来越严格的能源消耗和环境保护要求，汽车等燃油交通运输工具的轻量化已成为发展的必然趋势。镁合金是目前工业上得到有效应用的最轻的金属结构材料之一，镁密度仅为 1.74 g/cm^3，约为铝密度的 2/3，钛密度的 1/3，钢密度的 1/4，同时镁合金还具有较高的比强度和比刚度、优良的减振性、易于回收利用等一系列优点，广受材料和机械工作者的关注，被誉为“21 世纪重要的绿色工程金属结构材料”。此外，镁在地球上的储量丰富，约占地壳质量的 2.35%，其含量在金属元素中仅次于铝和铁含量。我国是世界上镁资源最为丰富的国家，约占全球镁资源总量的 50%以上。因此，合理地利用我国镁资源的优势、加快高性能镁合金研发、促进镁合金工程化并推广应用对我国具有重要战略意义[1−2]。

相比铝合金及合金钢，镁合金绝对强度低、塑性与成形性差、易腐蚀等基础性问题严重制约了镁合金材料的广泛应用。为解决这些难题，近十多年来国内外研究者们在提高镁合金强度、塑性、成形性和耐蚀性等方面做了大量的工作。在提升镁合金强度方面，大量的研究集中在使用时效析出相强化镁合金等方面[3−20]。例如，经时效析出强化后的 Mg-11Gd-4.5Y-1Nd-1.5Zn-0.5Zr(质量分数，%)镁合金抗拉强度可达 517 MPa，屈服强度达 482 MPa，但断裂伸长率仅 2%[21]。经时效后的 Mg-10Gd-5.7Y-1.6Zn-0.7Zr(质量分数，%)镁合金屈服强度可达 473 MPa，而断裂伸长率却不足 10%[22]。从这些研究可以看出，第二相析出(包括长程有序相)强化镁合金的效果比较显著，但合金的塑性也因第二相的产生而大幅度下降，使得析出强化的镁合金在工业上的应用受到限制。在提升镁合金塑性和耐蚀性方面，重庆大学潘复生团队开发的无熔剂纯净化技术制备的高纯度镁合金在提高塑性的同时，也显著提高了镁合金的耐蚀性[23−24]。在提升镁合金成形性方面，针对镁合金塑性变形特性和产业技术需求，重庆大学国家镁中心系统发展了镁合金新型非对称加工技术原理与体系，包括镁合金板材新型非对称挤压加工[25]、镁合金板材新型非对称改性[26]、镁合金棒材新型非对称制备加工[27]等，大大地提高了镁合金板材

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和棒材的室温成形性能。此外，重庆大学开发出了具有显著晶粒细化效果的镁基 Al_2Y、Al_2Nd、Al_2Ce 等细化剂，使镁合金铸态晶粒细化效率超过 80%，冶金质量显著改善，合金的成形性能显著提升[28−31]。这些研究工作在提升镁合金强度、或塑性、或成形性上取得了显著成效，但在解决镁合金高强度和高塑性(或成形性)的合理配合上仍不能令人满意。

在合金设计方面，如何在提升镁合金强度的同时不损害塑性或在提高塑性的同时不牺牲强度，已成为国内外新型镁合金发展中的研究热点和重点。研究工作主要集中在两个方面：一方面，重点研究合金元素对镁层错能的影响和基面/非基面滑移变化的微观表现；另一方面，重点研究基面滑移阻力的提高对启动非基面滑移的影响。对第一方面的内容，NIE Jian-feng、单智伟等做了大量杰出工作[32]，对后一方面的内容，重庆大学、西南大学等做了大量研究[33−34]。重庆大学潘复生团队等在研究合金元素对镁合金力学性能影响的过程中发现，某些特定原子固溶在镁中既能阻碍基面位错滑移，又能减小基面与非基面滑移阻力差异，使非基面滑移更容易开启，达到同时提高镁合金强度和塑性的目的，解决了镁合金塑性和强度的平衡优化问题。与国内外研究不同的是，这些发现重点突出了通过合金元素固溶缩小基面与非基面滑移阻力差距从而启动非基面滑移来提高镁合金塑性。针对这些发 现，潘复生团队提出了提高镁合金强度和塑性的合金设计新理论——“固溶强化增塑”(Solid solution strengthening and ductilizing, SSSD)，并应用该理论成功开发了多种新型高性能镁合金。

1　镁合金塑性变形的特点

镁及镁合金的晶体结构为典型的密排六方结构(Hexagonal close-packed，简称 HCP)，理想的镁金属由密排面按照……ABABABAB……这样的顺序循环堆垛而成，镁晶体的晶格常数为 a=3.21 Å，c=5.21 Å，轴比 c/a 为 1.623。镁合金与面心立方(奥氏体钢)和体心立方(铝合金、铁素体钢)金属的晶体结构不同，具有不同的滑移系与塑性变形机制[2]。

在镁合金中，基面〈a〉滑移是最容易开动的滑移系，室温下的临界剪切应力(Critical resolved shear stress，CRSS)约为其他非基面滑移的 1/100[35]，因此，在室温和低温下镁合金塑性变形以基面滑移为主，如图 1 所示为镁单晶在不同温度下的临界剪切应力[36]。升高温度有利于启动非基面滑移，但会导致晶粒明显粗化。在室温条件下，基面〈a〉滑移仅有两个独立滑移系，不满足 Von-Mises 屈服准则(即多晶体材料的均匀塑性变形至少需要 5 个独立的滑移系)，更远低于面心立方和体心立方晶格的 12 个独立滑移系，这导致镁合金室温和低温塑性变形能力较差[37]。此外，两个独立滑移系均在同一个滑移面，无法协调沿着 c 轴方向的变形，且基面滑移临界剪切应力远低于柱面和锥面的，导致镁合金后续变形加工过程中极易产生基面织构，均匀成形能力较差，性能均匀性严重恶化。因此，为了提高镁合金的均匀塑性变形能力，有必要通过对基面和非基面滑移阻力的调控，缩小基面与非基面滑

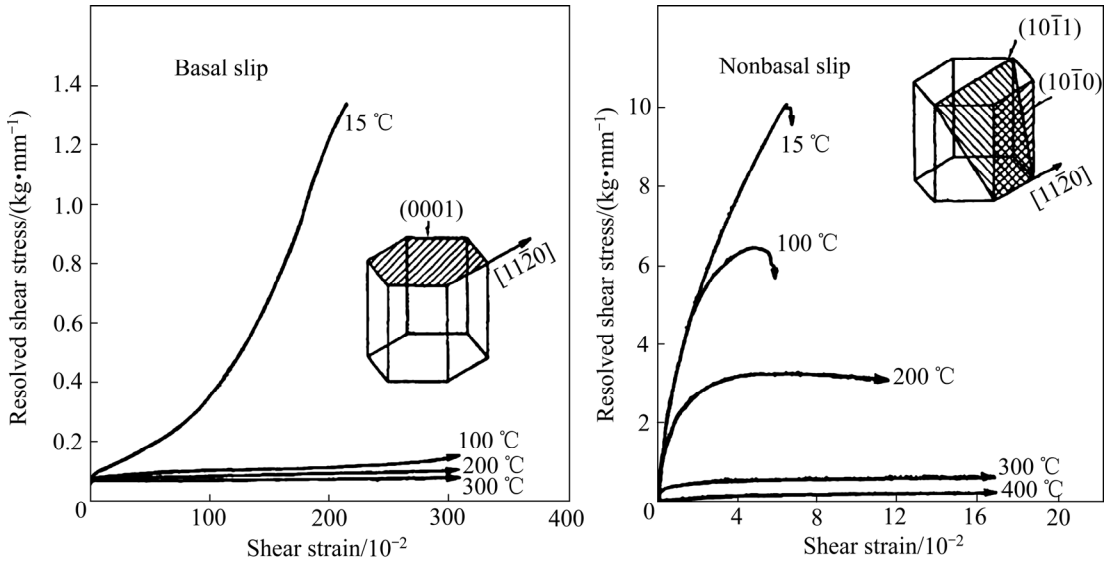

图 1　不同温度下镁单晶的基面与非基面滑移的临界剪切应力[36]

Fig. 1　CRSS for basal and non-basal slip systems in magnesium at different temperatures[36]

移阻力之间的差距，促进非基面滑移开启，从而改善镁合金的塑性成形能力。

2 “固溶强化增塑”合金设计理论的思路

在立方金属中，特别是在钢铁和铝合金材料中，固溶强化和析出强化一般都会降低材料塑性[38−39]。镁合金是六方晶体结构，析出强化降低塑性同样已得到大量的实验验证[4−7, 21, 40−41]，但对固溶强化的研究却发现了一些新现象。自 2002 年以来，重庆大学潘复生团队等针对合金元素影响镁合金强度和塑性的机理做了大量研究，如 Ag[42−43]、Nd[44−45]、Zn[46−47]、Mn[48−49]、Sn[32, 50]、Er[51−55]、Al[25, 56]、Y[57−58]、Ca[28, 59]、Ce[30, 60]、Li[31, 34]、Gd[61−62]、Sc[63]、Sr[64−65]等合金元素。在这些研究中，有关析出强化的研究都发现是合金的强度提高，塑性下降。重庆大学镁合金科研团队重点突出了元素固溶后基面与非基面滑移阻力变化如何影响塑性的研究。研究发现，某些特定元素原子固溶在镁中具有降低基面与非基面滑移阻力差异的独特作用，有利于非基面滑移的启动，进而提高镁合金的塑性。在 2005 年，潘复生及其合作者结合国内外研究工作提出了“固溶强化增塑”的合金设计思想，其主要思路见图 2 所示。

从图 2 可以看出，合金元素固溶在镁基体中，可增大或减小镁基面或非基面滑移阻力。当合金元素固溶后增大(或减小)基面和非基面的滑移阻力时，固溶体基面与非基面滑移阻力差值 $\Delta\tau'$ 和镁基面与非基面滑移阻力差值 $\Delta\tau$，有 $\Delta\tau'\approx\Delta\tau$、$\Delta\tau'>\Delta\tau$、$\Delta\tau'<\Delta\tau$ 3 种情况。当 $\Delta\tau'\approx\Delta\tau$ 时(见图 2(a)和(d))，元素固溶对基面与非基面滑移阻力差值的影响不大，并不能促进非基面滑移的开启，提高镁合金塑性；当 $\Delta\tau'>\Delta\tau$ 时(见图 2(c)，(f)，(g))，元素固溶增大了基面与非基面滑移阻力差值，使非基面滑移的启动更为困难，不利于镁合金塑性的改善；当 $\Delta\tau'<\Delta\tau$ 时(见图 2(b)，(e)，(h))，元素固溶减小了基面与非基面滑移阻力差值，有利于非基面滑移的开启，镁合金均匀塑性变形能力提高。而基面滑移阻力的减小不利于合金强度的提升，图 2(e)所示情况只能改善塑性，但会损失一定的强度，使得材料的工业应用受限。图 2(b)和图 2(h)所示为合金设计中应该追求的方向，其中图 2(h)所示条件最有利于同时提高强度和塑性。因此，当设计合金成分时，选用能够使基面滑移阻力增加且基面与非基面滑移阻力差值减小的合金元素，既可以产生强化

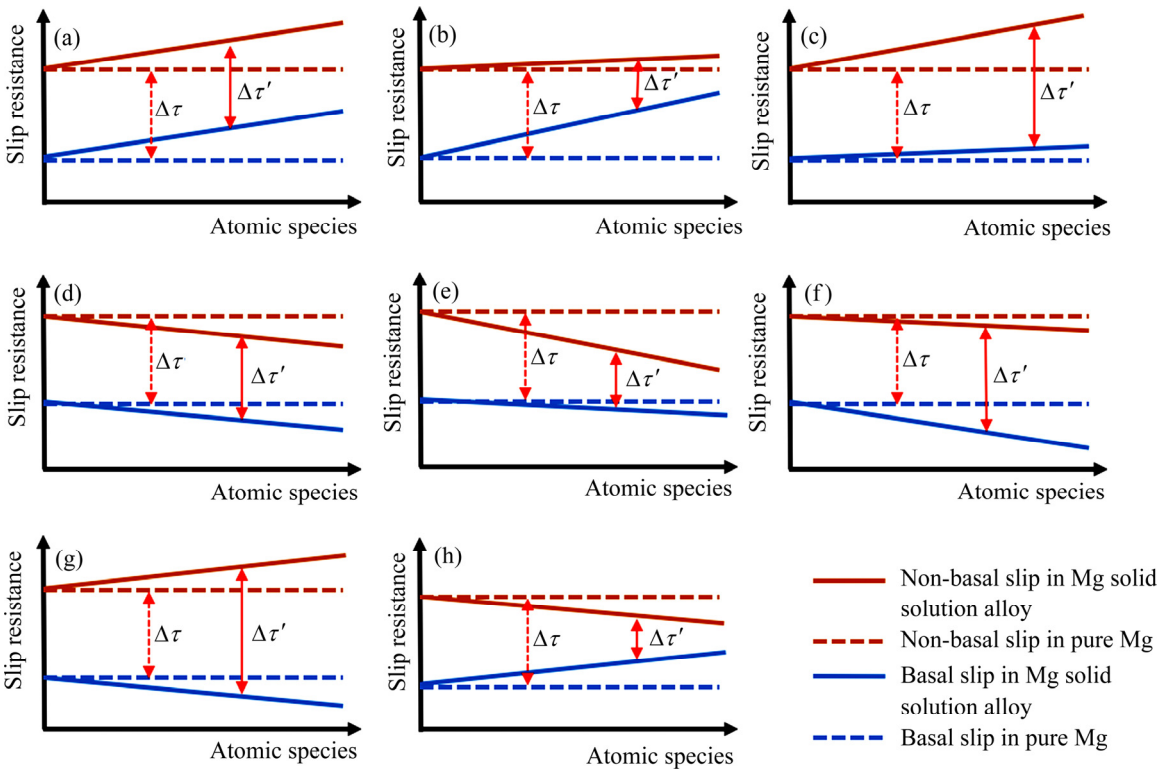

图 2 “固溶强化增塑”合金设计理论思路

Fig. 2 Thoughts of designing alloys by solid solution strengthening and ductilizing: (a), (d) $\Delta\tau'\approx\Delta\tau$; (b), (e), (h) $\Delta\tau'<\Delta\tau$; (c), (f), (g) $\Delta\tau'>\Delta\tau$

提高合金强度，又可以促进非基面滑移提高合金塑性，达到同时提高合金强度和塑性的效果，实现了“固溶强化增塑”的合金设计目的。

3 “固溶强化增塑”合金设计理论的理论计算

早期研究表明[66]，对密排六方结构的镁来说，c/a 轴比的变化将会改变原子间距，激发非基面滑移，从而提高镁合金的塑性。而后有越来越多的研究发现[67−70]，轴比的增大或减小与开启非基面滑移的难易程度没有对应关系，由此可见，轴比的变化并不是激发非基面滑移的关键因素。

层错是晶体中普遍存在的材料本征特征，层错能则是对应于特定的相对滑动位移所形成的层错所需要的能量。近年来，大量研究认为层错能的变化与基面、非基面滑移系的启动及合金塑性变形能力有关[71−76]。重庆大学潘复生团队研究了多种固溶原子对镁层错能的影响，研究发现合金元素 Al、Bi、Ca、Dy、Er、Ga、Gd、Ho、In、Lu、Nd、Pb、Sm、Sn、Y、Yb 可明显降低 I_1 层错能；合金元素 Ca、Dy、Er、Gd、Ho、Lu、Nd、Sm、Y、Yb 能大幅度降低柱面滑移系的非稳定层错能，有利于降低柱面位错滑移的临界分切应力；合金元素 Ag、Al、Ca、Dy、Er、Ga、Gd、Ho、Li、Lu、Nd、Sm、Y、Yb、Zn 有利于锥面滑移系的开启并提高镁合金的本征塑性[77−79]。

然而，这些有关层错能的研究都基于 0 K 下的第一性原理计算，为了更好地从理论上解释层错能与基面、非基面滑移系的启动及合金塑性变形能力之间的关系，有必要将 0 K 下对层错能的研究推广至有限温度。

潘复生团队[80]利用分子动力学模拟，采用次近邻修正嵌入原子势方法描述原子间的相互作用，研究了不同温度下多种合金元素对镁层错能的影响。对固溶含量为 0~3%(摩尔分数)的 Mg-Al、Mg-Zn 和 Mg-Y 这 3 种合金，研究了基面 $\langle 11\bar{2}0\rangle$、基面 $\langle 10\bar{1}0\rangle$、柱面 $\{10\bar{1}0\}\langle 11\bar{2}0\rangle$ 和锥面 $\{11\bar{2}2\}\langle 11\bar{2}3\rangle$ 4 个滑移系在 0~500 K 的温度下的广义层错能，重点探讨了固溶原子含量变化和温度对于层错能的影响，以及层错能与微观塑性变形模式的联系。为了更好地解释层错能对各滑移系开动情况的影响，对 MOITRA 等[81]提出的塑性成形参数 χ 进行了修正，定义了基面和柱面滑移开动相关的参数 χ_1 及基面和锥面滑移开动相关的参数 χ_2，来说明同一温度下层错能与镁合金塑性成形能力之间的关系，分别如式(1)和(2)所示：

$$\chi_1=\frac{(\gamma_{sf}^{B}/\gamma_{usf}^{B})_X}{(\gamma_{sf}^{B}/\gamma_{usf}^{B})_{Mg}}/\frac{(\gamma_{sf}^{M})_X}{(\gamma_{sf}^{M})_{Mg}} \tag{1}$$

$$\chi_2=\frac{(\gamma_{sf}^{B}/\gamma_{usf}^{B})_X}{(\gamma_{sf}^{B}/\gamma_{usf}^{B})_{Mg}}/\frac{(\gamma_{sf}^{Pyr}/\gamma_{usf}^{Pyr})_X}{(\gamma_{sf}^{Pyr}/\gamma_{usf}^{Pyr})_{Mg}} \tag{2}$$

式中：γ_{sf}^{B} 即基面 $\{0001\}\langle 10\bar{1}0\rangle$ 滑移系的稳定层错能；γ_{usf}^{B} 则是该滑移系中的不稳定层错能 γ_{usf}^{I}；γ_{sf}^{M} 为柱面滑移系稳定层错能和不稳定层错能的替代值；γ_{sf}^{Pyr} 和 γ_{usf}^{Pyr} 分别代表二级锥面上的稳定层错能 γ_{sf} 和不稳定层错能 γ_{usf}^{I}；角标的 Mg 代表纯 Mg 的情况，X 则代表含 Mg-X 合金的情况。由式(1)和(2)可知，纯镁的 χ_1 和 χ_2 值为 1，当 χ_1 和 χ_2 值大于 1 时，这时合金的塑性更好。图 3 所示为不同温度下 Mg-Al、Mg-Zn 和 Mg-Y 合金的 χ_1 和 χ_2 值随固溶原子增加的变化情况。可见 Mg-Al 合金的 χ_1 和 χ_2 值在各成分和温度下基本小于 1，且都随着 Al 的增加有轻微的下降。而 Mg-Zn 和 Mg-Y 合金的 χ_1 和 χ_2 值基本均大于 1，这两个合金的 χ_1 值都随着固溶原子的增加而上升，柱面滑移开动的倾向增加。χ_2 值则随着 Zn 原子含量增加有微弱地上升趋势。Mg-Y 中 χ_1 和 χ_2 值都随着 Y 的增加有明显的增加趋势，且都高于 Mg-Zn 的计算结果。根据 χ_1 和 χ_2 的计算结果可知，在同一温度下，固溶 Al 原子含量的增加并没有显著地改善镁合金非基面滑移开动的趋势，Zn 含量的增加有利于镁合金非基面滑移系的开动，而 Y 含量的提高则明显地增加了镁合金非基面滑移开动的可能性。

目前，大多数有关层错能变化对镁合金塑性影响的分析仅限于层错能增大或减小对位错滑移开启难易程度的影响。事实上，激活非基面滑移的根本是缩小基面与非基面滑移阻力的差距 $\Delta\tau$，而层错能与各滑移系滑移阻力有着直接关联。YASI 等[82]通过修正 Fleischer 模型建立了 0 K 时部分二元固溶体合金基面层错能与基面滑移阻力 CRSS 之间的表达关系，如式(3)所示[82]：

$$\Delta\tau_{crss(0001)}\approx(389\ \text{MPa})\{(\varepsilon_b/0.176)^2+(\varepsilon_{SFE}/0.176)^2-\varepsilon_b\varepsilon_{SFE}/2.98\}^{3/2}\cdot c_s^{1/2} \tag{3}$$

式中：$\Delta\tau_{crss(0001)}$ 表示与纯镁相比固溶合金基面 CRSS 的变化，MPa；ε_b 为固溶原子引起的尺寸错配；ε_{SFE} 为固溶原子引起的化学错配(即层错能变化)；c_s 为固溶原子浓度，%(摩尔分数)。此研究表明，层错能与固溶原子引起的尺寸错配能影响着基面滑移阻力 CRSS，

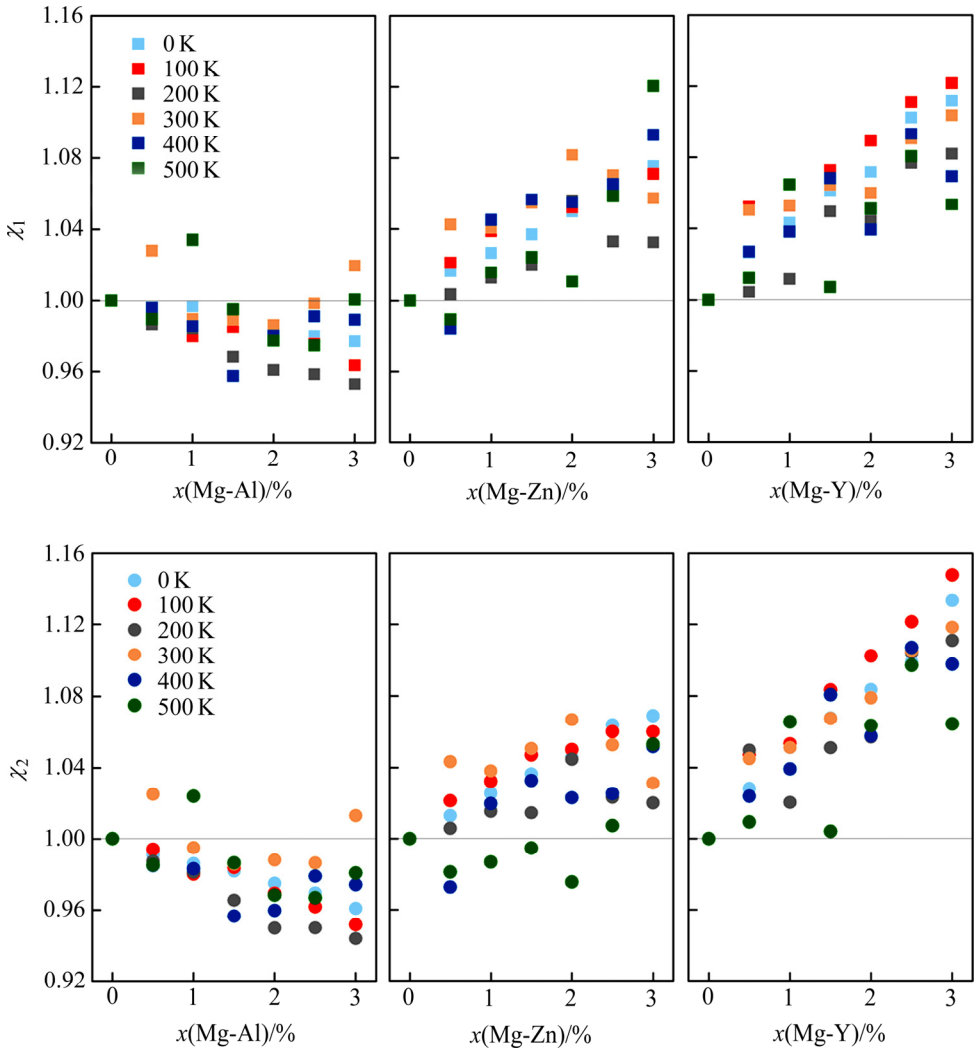

图 3 Mg-Al、Mg-Zn 和 Mg-Y 合金的 χ_1 和 χ_2 值随固溶原子含量增加的变化[80]

Fig. 3 Changing values of χ_1 and χ_2 with elevated solute concentration[80]

为定量分析层错能与滑移阻力之间的关系奠定了理论基础。在此基础上，作者应用式(3)计算了部分合金元素固溶后对镁基面滑移阻力 CRSS 变化的影响，包括合金元素在最大固溶度时基面滑移阻力 CRSS 变化 $\Delta\tau^{M}_{crss}$ 和固溶量为 1%(摩尔分数)时基面滑移阻力 CRSS 变化 $\Delta\tau^{1}_{crss}$，结果如表 1 所示。

从表 1 可知，Al、Zn、Y、Gd、Mn、Yb、Ag、Dy、Er 等元素均可增加镁基面滑移阻力，$\Delta\tau^{M}_{crss}$ 和 $\Delta\tau^{1}_{crss}$ 的值均为正值。比较 $\Delta\tau^{1}_{crss}$ 的值可知，相同含量下，Gd、Yb 和 Mn 增加镁基面滑移阻力的效果最佳，其次为 Y、Dy 和 Er。其中，Mn 是这些元素中成本最低的元素，对发展高塑性低成本镁合金非常有利，但 Mn 的固溶度较低，最大固溶度也只接近 1%(摩尔分数)。如何利用 Mn 的最大固溶度提高塑性和如何利用 Mn 的细小析出相细化晶粒进一步提高塑性是 Mn 在镁合金应用时必须同时考虑的问题。

从表 1 同样可以看出，对 Mg-Al、Mg-Zn 和 Mg-Y 合金而言，Al、Zn、Y 均可增加基面滑移阻力。对比 $\Delta\tau^{1}_{crss}$ 的值可知，在相同含量下，Y 增加镁基面滑移阻力的效果最佳，其次为 Zn、Al。由此可见，Al、Zn、Y 固溶在镁基体中基面滑移阻力增加，使得基面位错运动困难，产生基面的固溶强化，从而使合金的屈服强度提高。结合分子动力学模拟结果，Zn、Y 固溶有利于非基面滑移系的开启，提高合金的塑性。由此可见，Zn、Y 固溶在镁基体中可同时提高合金的屈服强度和塑性，即 Zn、Y 固溶实现了“固溶强化增塑”。有关 Zn、Y 固溶对非基面滑移阻力影响的详细结果待后续报道。

表 1　不同固溶元素 Mg 的基面滑移阻力 CRSS 变化 [1)]

Table 1　Basal CRSS changes of Mg with different solutes

Element	Maximum solid solubility, x_M/%	$\Delta\tau_{crss}^{M}$/MPa	$\Delta\tau_{crss}^{1}$/MPa
Al	11.5	7.08	2.09
Zn	2.69	5.23	3.19
Mn	0.996	10.13	10.13
Sc	15	6.11	1.58
Si	1.16	4.75	4.41
La	0.14	6.95	–
Sn	3.35	2.34	1.28
Y	3.4	16.71	9.06
Dy	4.83	18.89	8.6
Ag	3.83	8.54	4.37
Ti	0.12	0.49	–
Yb	1.2	11.15	10.18
Ca	0.44	6.85	–
Zr	1.04	1.00	0.98
Er	6.9	19.05	7.25
Fe	0.00043	0.24	–
Gd	4.53	23.22	10.91
Li	17	4.80	1.17

1) Unpublished work.

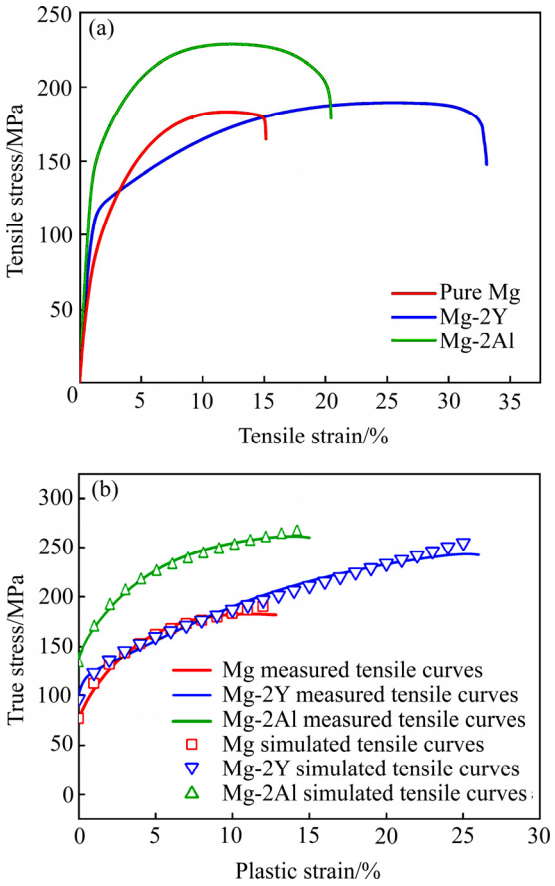

图 4　Mg、Mg-2Al、Mg-2Y 合金拉伸工程应力应变曲线(a)及真应力−应变拟合曲线(b)

Fig. 4　Tensile stress−strain curves(a) and ture stress−strain curves(b) of Mg, Mg-2Al, Mg-2Y alloys

4　“固溶强化增塑”合金设计理论的实验验证

“固溶强化增塑”合金设计理论可以采用应力应变曲线并结合粘塑性自洽(Visco-Plastic Self-Consistent)模型进行间接验证。图 4(a)所示为 Mg-X 二元系合金拉伸工程应力应变曲线，包括 Mg-2%Al、Mg-2%Y(质量分数)两个二元合金。由图 4(a)可见，添加 Al 和 Y 可明显提高纯镁的屈服强度和断裂伸长率，Mg-2Al 合金的强度最高，Mg-2Y 合金的断裂伸长率最高。

对纯 Mg、Mg-2Al、Mg-2Y 合金采用黏塑性自洽模型模拟了室温拉伸塑性变形过程(见图 4(b)所示)，并分析了 Al 和 Y 固溶对纯镁基面与非基面滑移阻力差值的影响，结果如表 2 所示。表 2 中 $\Delta\tau^{Basal}$、$\Delta\tau^{Prismatic}$、$\Delta\tau^{Pyramidal}$ 分别为 Al 和 Y 固溶后镁基面、柱面、锥面滑移阻力的变化，正值表示增加，负值表示减少；$\Delta\tau^{Pr-B}$ 和 $\Delta\tau^{Py-B}$ 分别代表合金柱面与基面滑移阻力差值和锥面与基面滑移阻力差值。当 Al 固溶到 Mg 中，基面滑移阻力 $\Delta\tau^{Basal}$ 增加，产生固溶强化，锥面与基面滑移阻力差值变化不大，对锥面滑移系开启的影响不明显，而柱面与基面滑移阻力差值增大，柱面滑移启动较为困难。当 Y 固溶到 Mg 中，基面滑移阻力 $\Delta\tau^{Basal}$ 小幅度增加，产生固溶强化，柱面与基面滑移阻力差值 $\Delta\tau^{Pr-B}$ 大幅下降，锥面与基面滑移阻力差值 $\Delta\tau^{Py-B}$ 明显下降，均有利于柱面和锥面滑移的启动，从而提高合金均匀塑性变形能力。进一步，Mg-2Al 合金基面滑移阻力增量较 Mg-2Y 合金大，因此 Mg-2Al 合金的屈服强度更高，而 Mg-2Y 合金的 $\Delta\tau^{Pr-B}$ 和 $\Delta\tau^{Py-B}$ 值比 Mg-2Al 合金的小很多，更有利于非基面滑移系的开启，故 Mg-2Y 合金比 Mg-2Al 合金的断裂伸长率高，塑性更好。由此可见，Al、Y 添加对纯 Mg 塑性变形能力的影响规律与分子动力学模拟计算结果相同。

表 3 所示为一些二元镁合金实验和文献报道的力学性能数据。从表 3 可以看出，添加 Al、Y、Mn、Gd、Zn、Er 等合金元素均可提高纯镁的抗拉强度、屈服强度和断裂伸长率。还可以看出，相同工艺下，随着 Gd、Zn 等合金元素的含量增加，二元合金的强度和断裂伸长率随之增加。尽管有晶粒细化、析出强化、合金纯化等方面的影响，但实验和文献报道的二元合

表 2 纯 Mg、Mg-2Al、Mg-2Y 合金力学性能及滑移阻力数据 1)

Table 2 Tensile mechanical properties and slip resistance of Mg, Mg-2Al and Mg-2Y alloys[1)]

Sample	Tensile yield strength /MPa	Elongation to failure /%	$\Delta\tau^{\mathrm{Basal}}$ / MPa	$\Delta\tau^{\mathrm{Prismatic}}$ / MPa	$\Delta\tau^{\mathrm{Pyramidal}}$ / MPa	$\Delta\tau^{\mathrm{Pr-B}}$ / MPa	$\Delta\tau^{\mathrm{Py-B}}$ / MPa
Mg	75.5	15.3	−	−	−	145	180
Mg-2Al	151.6	20.2	13	30	15	162	182
Mg-2Y	110.7	32.6	3	−90	−70	52	167

1) Recent work

表 3 一些二元合金力学拉伸学性能数据

Table 3 Tensile mechanical properties of Mg-X binary alloys

Alloy (mass fraction)	Tensile yield strength/MPa	Ultimate tensile strength/MPa	Elongation to failure/%	Processing route	Ref.
Mg	76	183	15.3	Heat treated at 420 ℃ for 24 h+extruded at 350 ℃	1)
Mg-2%Al	152	228	20.2	Heat treated at 420 ℃ for 24 h+extruded at 350 ℃	1)
Mg-3%Al	158	241	20.8	Heat treated at 420 ℃ for 24 h+extruded at 350 ℃	1)
Mg-4%Al	162	252	20.8	Heat treated at 420 ℃ for 24 h+extruded at 350 ℃	1)
Mg-2%Y	111	190	32.6	Heat treated at 520 ℃ for 24 h+extruded at 450 ℃	1)
Mg-2%Y	92	189	21	Heat treated at 480 ℃ for 12 h+extruded at 420 ℃	[83]
Mg-1.0%Mn	178	217	18.3	Heat treated at 500 ℃ for 24 h+extruded at 350 ℃	1)
Mg-0.89%Mn	204.3	234.1	38.8	As-extruded	[84]
Mg-1.0%Gd	80	186	25.5	Heat treated at 520 ℃ for 24 h+extruded at 450 ℃	1)
Mg-3.0%Gd	78	187	31.8	Heat treated at 520 ℃ for 24 h+extruded at 450 ℃	1)
Mg-0.75%Gd	145	210	12	Hot rolled at 400 ℃+annealed for 1 h at 380 ℃	[85]
Mg-2.75%Gd	160	205	21	Hot rolled at 400 ℃+annealed for 1 h at 380 ℃	[85]
Mg-4.65%Gd	165	210	26	Hot rolled at 400 ℃+annealed for 1 h at 380 ℃	[85]
Mg-1%Zn	126	215	17.3	Heat treated at 400 ℃ for 24 h+extruded at 350 ℃	1)
Mg-2%Zn	129	217	20.0	Heat treated at 400 ℃ for 24 h+extruded at 350 ℃	1)
Mg-4%Zn	139	242	25.1	Heat treated at 400 ℃ for 24 h+extruded at 350 ℃	1)
Mg-2%Er	83	251	19.6	Heat treated at 520 ℃ for 48 h+extruded at 400 ℃+annealed for 1 h at 400 ℃	[86]
Mg-4%Er	80	184	28.4	Heat treated at 520 ℃ for 48 h+extruded at 400 ℃+annealed for 20 min at 400 ℃	[86]
Mg-8%Er	153	260	44	As-extruded	[87]

1) Unpublished work

金变形和热处理工艺均在 400 ℃左右，体现了固溶元素对合金性能的影响，且从实验和文献报道数据均可看出合金元素固溶可同时提高纯镁的强度和塑性的趋势，进一步验证了合金元素在镁中的“固溶强化增塑”作用。

5 “固溶强化增塑”合金设计理论的应用

利用“固溶强化增塑”合金设计理论并结合长程有序相控制等途径，重庆大学镁合金科研团队开发了共二十多种新型高性能镁合金，包括超高强变形镁合金、超高强铸造镁合金、超高塑性镁合金、低成本高塑性镁合金、超轻合金、高电磁屏蔽性能镁合金、高导热性能镁合金等。图 5 所示分别为重庆大学镁合金科研团队开发的超高强变形镁合金[42](见图 5(a))、高塑性含锰镁合金[84](见图 5(b))、高强度高塑性铸造镁合金[88](见图 5(c))、超高塑性镁合金[89](见图 5(d))的拉伸力学性能曲线。此外，表 4 还列出了重庆大学镁合金科研团队开发的部分新型高性能镁合金及其性能。这些新开发的合金中，有 16 个已批准为国家标准牌号合金[90−91]，9 个已批准为国际标准牌号合金[92]。

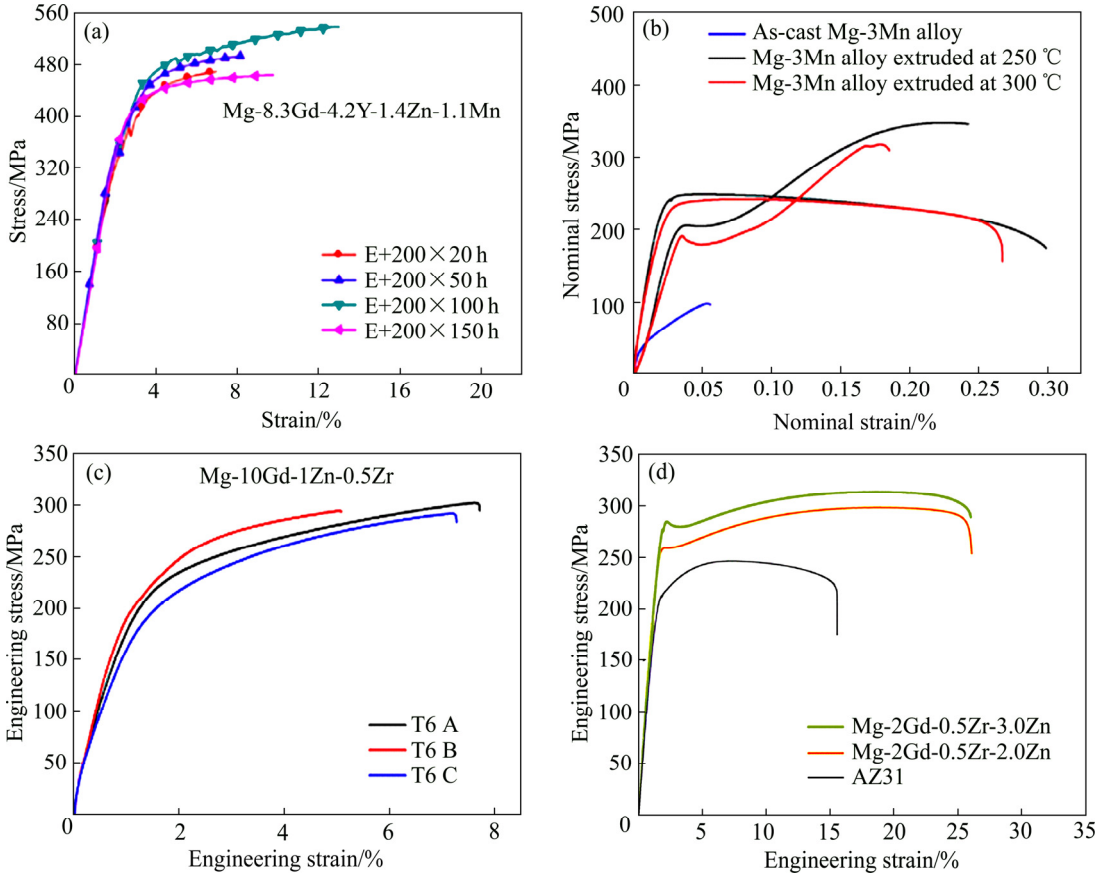

图 5　高性能镁合金的拉伸应力−应变曲线：(a) 超高强变形镁合金[42]；(b) 高塑性含锰镁合金[84]；(c) 高强度高塑性铸造镁合金[88]；(d) 超高塑性镁合金[89]

Fig. 5　Stress−strain curves of high properties magnesium alloys: (a) Ultra-high strength wrought magnesium alloys[42]; (b) High plastic wrought magnesium alloys[84]; (c) High strength and plastic cast magnesium alloys[88]; (d) Ultra-high plastic magnesium alloys[89]

表 4　重庆大学开发的部分新型高性能镁合金

Table 4　New high performance magnesium alloys developed by Chongqing University

Alloy	Series	Ultimate tensile strength/MPa	Elongation to failure/%	Ref.
Ultra-high plastic magnesium alloys	Mg-X-Gd	200−250	50−63	[89−92]
High plastic wrought magnesium alloys	Mg-Zn-Zr-Nd(Er)	230−300	20−40	[44, 51, 90]
Low-cost wrought magnesium alloys	Mg-Mn-Al	280−330	20−23	[90, 92, 93]
Wrought magnesium alloys without RE addition	Mg-Zn-Mn-Sn	380−400	8−10	[94]
Ultra-high strength wrought magnesium alloys	Mg-Gd-Y-Zn	500−550	10−13	[42, 90, 92]
High strength cast magnesium alloys	Mg-Gd-Y-Zn	330−380	9−12	[95]
Wrought magnesium alloys	Mg-Zn-Zr-Y-Ce	400−420	9−12	[96]
Ultra-light magnesium alloys	Mg-Li-Al-X	200−230	20−25	[97]

6　结语

如何提高镁合金室温塑性和低温热成形能力是镁合金推广应用中亟待解决的问题。镁合金“固溶强化增塑”理论可以为高塑性镁合金的开发提供一条合金设计的新思路，在实现强度提高的同时改善镁合金的塑性。在“固溶强化增塑”的合金设计思路中，Mn 的应用极有价值，一方面是因为 Mn 元素的成本极低，另一方面是因为 Mn 的低温固溶增塑效果非常显著并且有明显的析出效应，对发展低温成形的低成本超细

晶变形镁合金有重要意义。由于镁合金的阻尼性能也和位错的可动性密切相关，“固溶强化增塑”理论也可以为解决强度提高同时阻尼性能变差的矛盾提供新的解决思路，即可以通过固溶合金的设计尝试实现“固溶强化增阻”。另外，钛合金、铍合金、锌合金等六方晶体结构金属的材料塑性都比较差，用“固溶强化增塑”来提高塑性和成形性也是值得探索的工作。

“固溶强化增塑”原理在新型镁合金设计和开发中的准确应用还需要在多个方面进一步完善和发展。一是不同温度下合金元素影响非基面阻力的准确计算难度很大，特别是多个元素交互作用下计算难度更大；二是实验验证的有效性和准确性有待进一步改善；三是缺乏大量准确的镁合金多元相图，合金元素固溶量变化目前并不完全清楚，大量热力学和动力学研究还亟待加强；四是在镁合金固溶、析出和位错等方面研究中，热力学、动力学研究和微观组织精准研究脱节现象依然严重，协同研究非常重要。

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Development and application of “solid solution strengthening and ductilizing” for magnesium alloys

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Abstract: Mg has a HCP structure with few slip systems, and the slip resistance for basal slip is much lower than that of prismatic and pyramidal slip. It is generally difficult to activate non-basal slip at room and low temperature, resulting in a poor ductility and formability of magnesium alloys. Chongqing University and other institutions have found that the solution of certain elements in Mg can not only improve the strength by hindering the basal slip, but also improve the ductility by reducing the gap of slip resistance between basal slip and non-basal slip. According to these results, Chongqing University has proposed a theory of alloy design —— solid solution strengthening and ductilizing (SSSD) for magnesium alloy, which has become a new way to optimize the strength-ductility balance of magnesium alloys in the past decade. Using SSSD theory, Chongqing University has developed many new high performance magnesium alloys, among which over 10 alloys have been included in GB/T national standard and ISO international standard.

Key words: magnesium alloys; strength; ductility; solid solution strengthening and ductilizing

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