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1 **The influence of shearable and nonshearable precipitates on the**
2 **Portevin-Le Chatelier behavior in precipitation hardening**
3 **AlMgScZr alloys**

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16
17 **Abstract:** The well-known mechanisms of interaction between precipitates and
18 dislocations are shearing (for shearable precipitates) and bypassing mechanisms (for
19 nonshearable precipitates). The transition from shearable to nonshearable precipitates
20 in precipitation hardening alloys leads to changes of dislocation motion mode and
21 dislocation multiplication behavior, which inevitably causes different PLC behaviors.
22 In this study, we systematically investigate the influence of shearable and
23 nonshearable Al₃(Sc, Zr) precipitates on PLC behaviors by experimental
24 characterization for precipitation hardening AlMgScZr alloys. We analyze the onset
25 strain, critical strain-rate range, serration amplitude, and propagation behavior of PLC
26 bands in detail for AlMgScZr alloys with shearable and nonshearable precipitates,
27 respectively. We find that the transition from shearable to nonshearable precipitates
28 changes the PLC band propagation behavior, decreases the magnitude of serration
29 amplitude, expands the strain-rate range as well as decreases the critical strain rate
30 between normal behavior (the critical strain increases with the increase of strain rate)

31 and inverse behavior (the critical strain decreases with the increase of strain rate)
32 regimes due to the different dislocation-precipitate interactions. Besides, the transition
33 from shearable to nonshearable precipitates increases the onset strain at normal
34 behavior while decreases the onset strain at inverse behavior depending on the
35 different roles of precipitates in the solute-dislocation interaction. Finally, we reveal
36 the nature of influence of different dislocation-precipitate interactions on PLC
37 behavior considering different strengthening mechanisms based on quantitative
38 characterization on precipitates and dislocation density.

39

40 **Keywords: Aluminium alloy, Precipitates, Portevin-Le Chatelier effect,**
41 **Strengthening mechanism, Dislocations**

42

43 **1. Introduction**

44 Portevin-Le Chatelier (PLC) is a kind of dynamic strain ageing (DSA)
45 phenomenon due to the interaction between diffusing solute atoms and mobile
46 dislocations, which has been found in many alloy systems, particularly in AlMg alloys
47 [McCormick, 1972a; Van den Beukel, 1975; Estrin and Kubin, 1990; Kubin and
48 Estrin, 1990; Robinson, 1994; Hähner, 1997; Lebyodkin et al., 2000; Picu, 2004;
49 Fressengeas et al., 2005; Antolovich and Armstrong, 2014; Aboufadel et al., 2015;
50 Swaminathan et al., 2015; Zhang et al., 2017; Tsai et al., 2019]. On the atomic level,
51 the PLC phenomenon is caused by the diffusion of solute atoms [Cottrell and Bilby,
52 1949; Yoshinaga and Morozumi, 1971a, 1971b; Aifantis, 1987; Estrin and Kubin,
53 1990; Picu, 2004; Curtin et al., 2006; Soare and Curtin, 2008a, 2008b; Aboufadel et al.,
54 2015]. The dislocations are efficiently pinned by solutes in direct relation with the
55 solute which has segregated to its core. The longer the waiting time (i.e. the lower the
56 imposed strain rate) of a dislocation between two jumps, the higher the solute
57 concentration at dislocation core and therefore the higher the resulting flow stress
58 [McCormick, 1972a; Estrin and Kubin, 1990, 1991, 1995; Kubin and Estrin, 1985,
59 1990; Curtin et al., 2006]. The interaction between dislocations and solute atoms can
60 result in a decrease of the flow stress associated to an increase of the imposed strain
61 rate and thus to a negative strain rate sensitivity (SRS) [Van den Beukel, 1975;
62 Mulford and Kocks, 1979; Louat, 1981; Kubin et al., 1988; Estrin and Kubin, 1990,

63 1995; Kubin and Estrin, 1990; Picu, 2004; Curtin et al., 2006]. On a macroscopic
64 scale, the PLC phenomenon is generally manifested by sudden localizations of the
65 plastic strain in transient deformation bands, leading to jerky flow features on the
66 stress strain curves [Zhang et al., 2001; Kok et al., 2003; Zhang et al., 2005; Benallal
67 et al., 2008; Zavattieri et al., 2009; Manach et al., 2014; Klusemann et al., 2015;
68 Shibkov et al., 2016; Yuzbekova et al., 2017; Yang et al., 2018; Zhemchuzhnikova et
69 al., 2018; Ren et al., 2021]. When PLC occurs, each serration of the deformation
70 curve corresponds to the development of localized deformation bands [Klusemann et
71 al., 2015; Yuzbekova et al., 2017; Zhemchuzhnikova et al., 2018]. The nucleation and
72 propagation of deformation bands have been proved to have a close relation to jerky
73 flow for many materials with PLC behavior, especially in AlMg alloys [Zhang et al.,
74 2001; Ait-Amokhtar et al., 2006a, 2006b; Zavattieri et al., 2009; Manach et al., 2014;
75 Klusemann et al., 2015; Yang et al., 2018]. Three major types of PLC bands, namely
76 continuous propagating type A, hopping type B and non-propagating type C bands are
77 defined based on the serrated curves [Pink and Grinberg, 1981; Robinson and Shaw,
78 1994; Jiang et al., 2007; Yilmaz, 2011; Hu et al., 2012; Zhang et al., 2017].

79 The PLC bands generated during deformation leave undesirable traces on the
80 surface of sheet products during deep drawing restricting the commercial application
81 of some alloys [Dierke et al., 2007; Böhlke et al., 2009; Yilmaz, 2011; Duan et al.,
82 2021]. Moreover, the ductility will be decreased due to the appearance of PLC [Kang
83 et al., 2006; Aretz, 2007; Halim et al., 2007; Zhang et al., 2012; Alinaghian et al.,
84 2014; Keralavarma et al., 2014; Rousselier and Quilici, 2015]. Undoubtedly, the PLC
85 effect is expected to be diminished by designing specific alloy compositions or by
86 controlling processing conditions. Nanostructured precipitates are beneficial to
87 eliminate the PLC effect, which already has been reported previously [Riley and
88 McCormick, 1977; Hayes and Hayes, 1982, 1984; Wert and Wycliffe, 1985; Pink,
89 1989; Kumar and McShane, 1993; Kumar et al., 1996; Thevenet et al., 1999;
90 Nalawade et al., 2008; Wang et al., 2016; Cai et al., 2017]. Pink and Thevenet et al.
91 [Pink, 1989; Thevenet et al., 1999] proved that the jerky flow in AlZnMg alloy would
92 be restrained by precipitates, especially the appearance of GP-zones could increase
93 the onset strain of PLC significantly. Kumar and McShane [Kumar and McShane,
94 1993] stated that the transformation of metastable precipitates (from coherent to
95 incoherent state) in the process of ageing can lead to the disappearance of PLC in
96 AlLi alloys. However, some scholars underlined that solute concentration in the

97 matrix decreased due to precipitation, which suppressed PLC [Riley and McCormick,
98 1977; Wert and Wycliffe, 1985; Nalawade et al., 2008]. There are others who
99 emphasized that dislocations shearing of coherent precipitates directly resulted in
100 serrations on the flow stress curve of AlLi alloys [Kumar and Pink, 1994, 1997;
101 Kumar, 1995; Pink et al., 2000]. While some people demonstrated that dislocations
102 shearing precipitates would not lead to jerky flow in the AlLi or in the Ni-based
103 superalloy [Chmelík et al., 1998; Wang et al., 2016]. In the mentioned studies [Riley
104 and McCormick, 1977; Hayes and Hayes, 1982, 1984; Wert and Wycliffe, 1985; Pink,
105 1989; Kumar and McShane, 1993; Kumar and Pink, 1994; Kumar et al., 1995; Kumar
106 et al., 1996; Chmelík et al., 1998; Thevenet et al., 1999; Nalawade et al., 2008; Wang
107 et al., 2016; Cai et al., 2017], the solute concentration in the matrix decreases when
108 the precipitates appear, so it is difficult to distinguish whether the change of PLC
109 behavior is caused by precipitates or solute concentration variations. Although the
110 aforementioned works have focused on the effect of nanosized precipitates on the
111 PLC behavior, they do not focus on the influence of precipitates on the propagation
112 behavior of PLC bands. Only recently, Cai et al. and Wang et al. indicated that type A
113 bands can turn into type B, and finally became type C with increasing precipitate
114 volume fraction in Ni-based alloys [Cai et al., 2015, 2017; Wang et al., 2019]. Cai et
115 al. explained that these changes were due to a change in the plastic deformation
116 mechanism, which turned from dislocation slip to stacking fault generation [Cai et al.,
117 2017]. Obviously, this mechanism cannot be applied in the AlMg system of the
118 present study. On the other hand, to the authors' knowledge, no attention is paid on
119 the influence of precipitate size (i.e. the effect of different precipitate-dislocation
120 interactions) on PLC behavior without considering the variation of solute
121 concentration. This impedes drawing clear conclusions about the influence of
122 precipitates on the PLC behavior.

123 The well-known mechanisms between precipitates and dislocations are shearing
124 (for shearable precipitates) and bypassing mechanisms (for nonshearable precipitates)
125 [Ardell, 1985; Gladman, 1999; Knipling et al., 2010, 2011; Hull and Bacon, 2011;
126 Fan et al., 2018; Ming et al., 2018; Barnett et al., 2019; Xu et al., 2019; Chen et al.,
127 2021]. The influence of precipitates on the PLC behavior lies in revealing the
128 mechanisms of appearance of precipitates on existing dislocation-dislocation and
129 dislocation-solute interactions. In view of current problems, we design the AlMgScZr

130 alloy system with the $\text{Al}_3(\text{Sc}, \text{Zr})$ precipitates on AlMg matrix to quantitatively study
131 the effects of shearable and nonshearable precipitates on PLC behavior with the aid of
132 experimental characterization in this study. The advantage of this alloy is that the
133 solute Mg concentration keeps unchanged when $\text{Al}_3(\text{Sc}, \text{Zr})$ precipitates occur, which
134 avoids the influence of solute concentration variation on PLC (It is because the
135 interaction of Mg atoms and dislocations is the one which leads to PLC phenomenon).
136 The shearable and nonshearable $\text{Al}_3(\text{Sc}, \text{Zr})$ precipitates can be obtained through
137 controlling heat treatment, and their quantitative characterization can be easily
138 obtained by transmission electron microscopy (TEM) and small angle neutron
139 scattering (SANS) techniques due to its good uniform dispersion, spherical shape and
140 thermal stability [Fuller et al., 2003; Clouet et al., 2006; Voorhees, 2006; Deschamps
141 et al., 2007; Deschamps and Geuser, 2011; Chen et al., 2021]. In this study, we focus
142 our attention on effects of different dislocation-precipitate interactions (shearing and
143 bypassing mechanisms) on PLC behavior including propagation behavior of PLC
144 band without considering the variation of solute concentration, which has not been
145 reported to our knowledge. We will demonstrate the reason of the influence of
146 dislocation-precipitate interaction on PLC behavior with considering different
147 strengthening mechanisms based on the quantitative characterization on dislocation
148 density measured by synchrotron radiation X-ray diffraction.

149

150 **2. Materials and methods**

151 Two cast AlMg and AlMgScZr alloys were used. The chemical compositions in
152 wt.% of AlMg and AlMgScZr are Al-5.7%Mg and Al-5.4%Mg-0.2%Sc-0.15%Zr,
153 respectively. The AlMg alloy was first homogenized at 813 K for 24 h and then
154 extruded at 623 K with a ratio of 10:1. It is known that $\text{Al}_3(\text{Sc}, \text{Zr})$ precipitates
155 nucleate at aging temperatures higher than 598 K [Knipling et al., 2010, 2011], for
156 which reason the cast AlMgScZr alloys were first aged at 623 K, 773 K for 3h and at
157 813 K for 24h correspondingly, with the purpose of obtaining shearable and
158 nonshearable $\text{Al}_3(\text{Sc}, \text{Zr})$ precipitates. After aging, the AlMgScZr alloys were also
159 extruded at 623 K with an extrusion ratio of 10:1.

160 TEM samples were extracted from the AlMg and AlMgScZr extrusion specimens,
161 which were polished mechanically, followed by electrochemical polishing by using a
162 twin-jet polishing unit with a mixed solution of 1/3 methanol and 2/3 nitric acid. TEM

163 experiments were carried out using a high-resolution Cs-corrected FEI Themis
164 microscope operating at 300 kV and a JOEM-2100F operating at 200 kV. The
165 quantitative characterization on precipitates is necessary in the discussion of different
166 dislocation-precipitate interactions on PLC behaviors, which was obtained by SANS.
167 The SANS experiment was performed on the SANS instrument at China Spallation
168 Neutron Source (CSNS) [Ke et al., 2018]. The detailed information about SANS and
169 data fitting process can be found in reference [Chen et al., 2021].

170 The tensile deformation of heat-treated AlMg and AlMgScZr alloys was carried
171 out at room temperature (298 K) using strain rates between $5 \times 10^{-1} - 1.0 \times 10^{-6} \text{ s}^{-1}$. The
172 gauge section of tensile specimens with dog-bone shape was $15 \times 3 \times 2 \text{ mm}^3$. The
173 dislocation densities of extruded AlMg and AlMgScZr alloys were determined by
174 high resolution X-ray diffraction line profile analysis (XLPA), which were measured
175 by synchrotron radiation X-ray diffraction (SRXRD) conducted at BL14B1 beamline
176 in Shanghai Synchrotron Radiation Facility (SSRF). The average dislocation density
177 was determined from the asymptotic behavior of the second and fourth-order
178 restricted moments [Borbély and Groma, 2001] of the intensity distribution
179 corresponding to the 200, 220 and 311 diffraction peaks of the samples. the
180 dislocation density in this study is determined by variance method. The asymptotic
181 behavior of the variance in our method does not depend on dislocation arrangement
182 [Kalácska et al., 2017]. This is because the asymptotic behavior is only influenced by
183 the strain field in the vicinity of dislocations and not by the dislocation arrangement
184 (dislocations far away). This is a well-known property of the Fourier transform
185 linking small distances in the real space (close to dislocations) to large distances in the
186 wave vector space (the tails of the diffraction peak). Therefore, the determined total
187 dislocation density by SRXRD is reliable, although the distribution of dislocation is
188 non-uniform in AlMgScZr-623K/3h alloy (seen discussion section). The detailed
189 information about variance method is shown in supplementary materials.

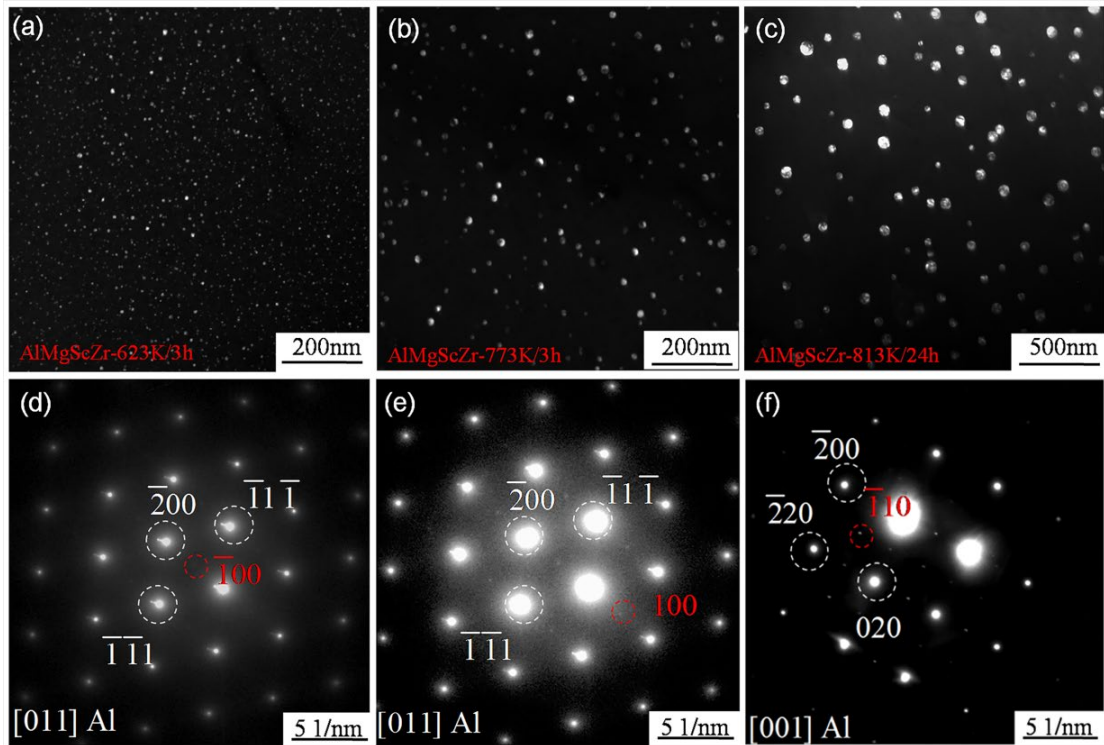
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191 **3. Results and analysis**

192 **3.1 Quantitative characterization of Al₃(Sc, Zr) precipitates**

193 Fig. 1 shows dark-field TEM images of Al₃(Sc, Zr) precipitates and the
194 corresponding selected area electron diffraction (SAED) patterns for

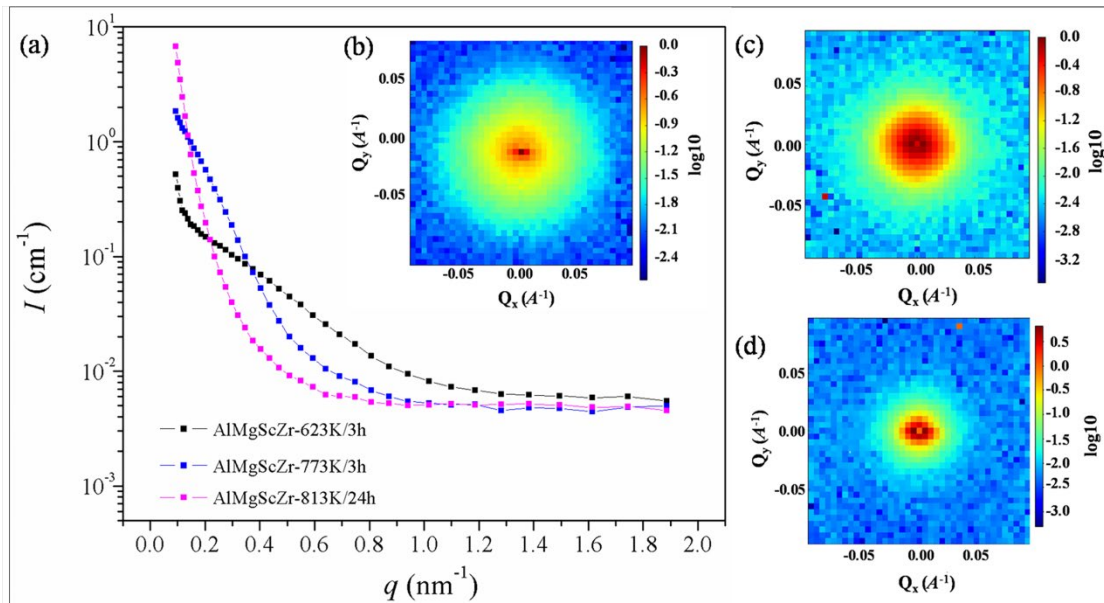
195 AlMgScZr-623K/3h, AlMgScZr-773K/3h and AlMgScZr-813K/24h alloys. The
196 precipitates have a spherical shape and are distributed relatively uniformly in the
197 aluminum matrix (Fig. 1). Fig. 2 shows the curves of scattering intensity I versus
198 scattering vector q captured by SANS and the corresponding raw two-dimensional
199 scattering data for these three kinds of AlMgScZr alloys. The mean radius, volume
200 fraction and precipitate inter-distance obtained from SANS are shown in Table 1. The
201 results in Table 1 reveal that precipitate size increases with increasing aging
202 temperature and mean radius is about 2.5, 10.2 and 22.6 nm for AlMgScZr-623K/3h,
203 AlMgScZr-773K/3h, AlMgScZr-813K/24h alloys. The corresponding volume fraction
204 of these three samples is about 0.46%, 0.42% and 0.40%, which remains practically
205 unchanged considering the error of the evaluation of about 10%. The results also
206 show that the increase of precipitate size leads to an increase of inter-precipitate
207 distance (Table 1) [Knipling et al., 2010, 2011]. The quantitative information on
208 Al₃(Sc, Zr) (average radius and volume fraction) obtained by TEM and SANS is close
209 to the previous results for the Al-Sc-Zr systems [Seidman et al., 2002; Fuller et al.,
210 2003; Clouet et al., 2005; Marquis and Seidman, 2005; Deschamps et al., 2007;
211 Knipling et al., 2010; Taendl et al., 2016]. It can be seen from Table 1 that the
212 precipitate radius for AlMgScZr-623K/3h alloy is smaller than 3 nm (the critical
213 radius between shearable and nonshearable precipitates is about 3-4 nm [Fazeli et al.,
214 2008; Knipling et al., 2010; Lai et al., 2013; Okle et al., 2019; Chen et al., 2021], thus
215 the precipitates in this alloy are defined as shearable precipitates. Analogously, the
216 precipitates in AlMgScZr-773K/3h and AlMgScZr-813K/24h alloys are defined as
217 nonshearable precipitates. In order to verify that the above definition on precipitates
218 in the three AlMgScZr alloys are reasonable, their strengthening effect are quantified
219 by comparing model prediction and experimental results, as discussed in 4.1 section.
220 The model prediction and experimental results are in good agreement within
221 experimental accuracy, further demonstrating the precipitates in AlMgScZr-623K/3h
222 alloy are shearable as well as in AlMgScZr-773K/3h and AlMgScZr-813K/3h alloys
223 are nonshearable.



224

225 Fig. 1 Characterization on $\text{Al}_3(\text{Sc}, \text{Zr})$ precipitates by dark-field TEM. The dark-field
 226 TEM images of $\text{Al}_3(\text{Sc}, \text{Zr})$ precipitates of (a) AlMgScZr-623K/3h alloy, (b)
 227 AlMgScZr-773K/3h alloy and (c) AlMgScZr-813K/24h alloy; (d-f) the corresponding
 228 SAED patterns of (a-c). The extracted spots for above dark-field images (a-c) are
 229 $(\bar{1}00)$, (100) and $(\bar{1}10)$, respectively.

230



231

232 Fig. 2 Quantitative characterization on precipitates by SANS. (a) Scattering intensity I
 233 versus scattering vector q for AlMgScZr-623K/3h [Chen et al., 2021],
 234 AlMgScZr-773K/3h [Chen et al., 2021] and AlMgScZr-813K/24h alloys; (b-d) the
 235 corresponding raw two-dimensional scattering data for AlMgScZr-623K/3h,

236 AlMgScZr-773K/3h and AlMgScZr-813K/24h alloys. Here figure (b) is inserted in
 237 figure (a).

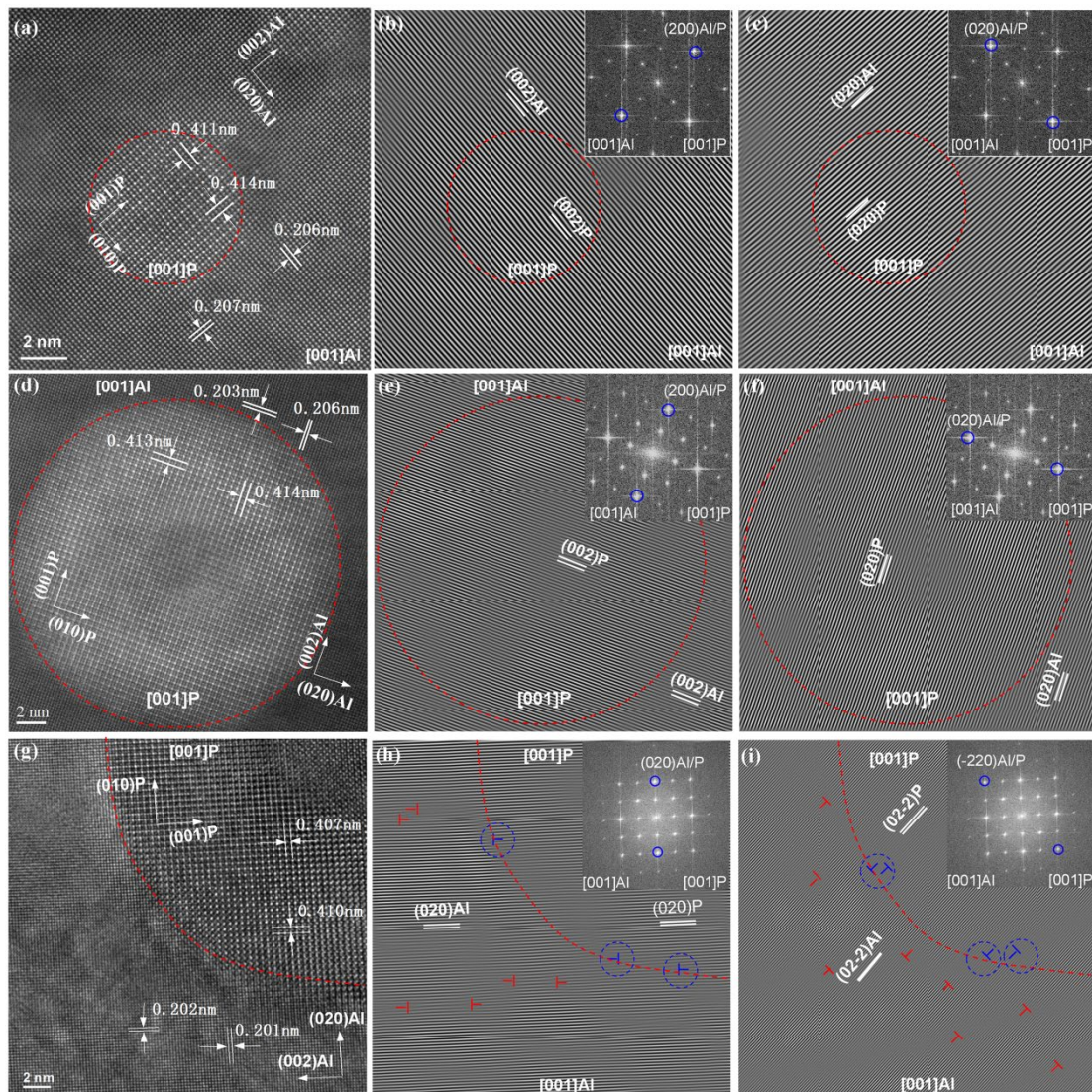
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239 Table 1 Quantitative information on precipitates provided by SANS and precipitate attribute. The
 240 critical precipitate size between shearable and nonshearable precipitates is about 3-4 nm. When
 241 the radius of precipitates is smaller than 3 nm, the precipitates are defined shearable precipitates.
 242 When the radius of precipitates is higher than 4 nm, the precipitates are defined as nonshearable
 243 precipitates.

Samples	Average radius (nm)	Volume fraction (%)	Inter-distance (nm)	Precipitate attribute
AlMg	-	-	-	without precipitates
AlMgScZr-623K/3h	2.5 ± 0.7	0.46 ± 0.05	49.4	shearable precipitates
AlMgScZr-773K/3h	10.2 ± 2.2	0.42 ± 0.04	211.7	nonshearable precipitates
AlMgScZr-813K/24h	22.6 ± 4.9	0.40 ± 0.03	481.5	nonshearable precipitates

244

245



246

247 Fig. 3 HRTEM and IFFT images of shearable and nonshearable $\text{Al}_3(\text{Sc}, \text{Zr})$
248 precipitates: (a-c) shearable precipitates (in the AlMgScZr-623K/3h sample); (d-f)
249 nonshearable and coherent precipitates (in the AlMgScZr-773K/3h sample); (g-i)
250 nonshearable and semi-coherent precipitates (in the AlMgScZr-813K/24h sample). (a),
251 (d) and (g) are HRTEM images, (b-c), (e-f) and (h-i) are corresponding IFFT images.
252 The red dashed lines represent the boundary between aluminum matrix and
253 precipitates. The red "T" denotes dislocations in the aluminum matrix, while the blue
254 "T" refers to interfacial dislocations. No interfacial dislocations are observed for
255 AlMgScZr-623K/3h and AlMgScZr-773K/3h samples (coherent precipitates) and lots
256 of interfacial dislocations are observed for AlMgScZr-813K/24h sample
257 (semi-coherent precipitates). The (002)Al, (020)Al, (001)P, (010)P refers to the (002),
258 (020) crystallographic planes of aluminum matrix and (001), (010) crystallographic
259 planes of the precipitates, respectively; [001]P and [001]Al represent the crystal zone
260 axis for the precipitate and matrix, respectively.

261

262 Figs. 3(a), (d) and (g) show high resolution TEM (HRTEM) images of shearable
263 (AlMgScZr-623K/3h alloy) and nonshearable $\text{Al}_3(\text{Sc}, \text{Zr})$ precipitates
264 (AlMgScZr-773K/3h and AlMgScZr-813K/24 alloys). Their Inverse Fast Fourier
265 Transform (IFFT) reveals that the Al/ $\text{Al}_3(\text{Sc}, \text{Zr})$ interface is free of dislocations for
266 the shearable precipitates (Figs. 3(b-c)) and nonshearable precipitates of
267 AlMgScZr-773K/3h alloy (Figs. 3(e-f)), manifesting that the precipitates are coherent
268 with aluminum matrix. However, lots of interfacial dislocations are observed for the
269 nonshearable precipitates of AlMgScZr-813K/24 alloy (Figs. 3(h-i)) demonstrating
270 that the precipitates are semi-coherent with aluminum matrix. According to previous
271 literature [Iwamura and Miura, 2004], the driving force required for the formation of
272 interfacial dislocations mainly results from the elastic energy caused by the lattice
273 misfit. Hence, the appearance of interfacial dislocations indicates loss of coherency
274 [Marquis and Seidman, 2001]. It has been shown for the case of Al-Sc alloys that
275 precipitates are coherent with the aluminum matrix when their average radius $r_{\text{ave}} < 15$
276 nm, and become semi-coherent when $r_{\text{ave}} > 40$ nm [Iwamura and Miura, 2004]. In the
277 intermediate stage ($15 < r_{\text{ave}} < 40$ nm), there are coexisting coherent and semi-coherent
278 precipitates¹[Iwamura and Miura, 2004]. Therefore, in this study, precipitates are
279 found to be coherent completely with matrix for AlMgScZr-623K/3h and
280 AlMgScZr-773K/3h alloys, but the coherent and semi-coherent precipitates coexist

¹ Note: For AlMgScZr alloy, the additions of Mg and Zr elements may increase the critical radius between coherent and semi-coherent precipitate due to Mg and Zr elements reducing the lattice mismatch between the matrix and precipitate.

281 for AlMgScZr-813K/24h alloy.

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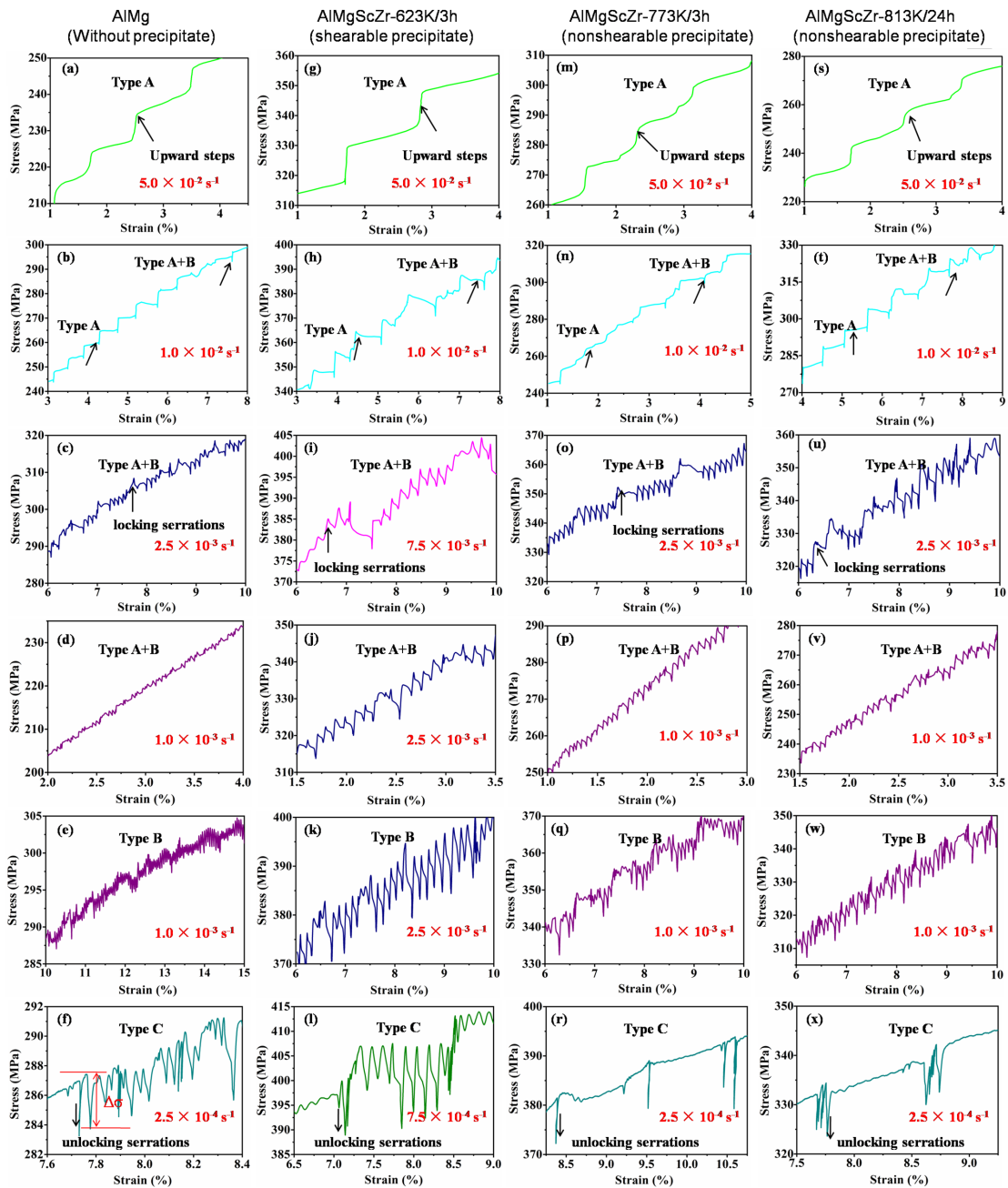
283 3.2 Deformation curves of AlMg and AlMgScZr alloys

284 Fig. 4 shows the engineering stress strain curves of the AlMg and three kinds of
285 AlMgScZr alloys deformed at different strain rates. For the AlMg alloy, the PLC
286 phenomenon occurs in a large interval of strain rate spanning from $1.0 \times 10^{-1} \text{ s}^{-1}$ to
287 $1.0 \times 10^{-6} \text{ s}^{-1}$ (Fig. 4(a)). The deformation curves of AlMgScZr-623K/3h,
288 AlMgScZr-773K/3h and AlMgScZr-813K/24h alloys are displayed in Figs. 4(b), (c)
289 and (d). The corresponding strain-rate ranges for the occurrence of PLC phenomenon
290 are $1.0 \times 10^{-1} - 2.5 \times 10^{-4} \text{ s}^{-1}$, $1.0 \times 10^{-1} - 1.0 \times 10^{-5} \text{ s}^{-1}$ and $1.0 \times 10^{-1} - 1.0 \times 10^{-6} \text{ s}^{-1}$. Fig. 5
291 shows the magnified figures of serrated flow of AlMg and three AlMgScZr alloys.
292 From Fig. 5, the PLC phenomenon is manifested by approximately upward stress
293 steps, followed by a stress plateau when strain rate is higher than $1.0 \times 10^{-2} \text{ s}^{-1}$, which
294 is similar to the previous studies [Böhlke et al., 2009; Duan et al., 2021]. When strain
295 rate is less than $1.0 \times 10^{-2} \text{ s}^{-1}$, the PLC appears like saw-like serrations. Type A, B and
296 C PLC bands are identified from the stress strain curves, and they appear at high,
297 middle and low strain rates respectively (the mixed type A+B is also observed) both
298 for AlMg and three AlMgScZr alloys. Upward steps, locking serrations and unlocking
299 serrations are visible in magnified regions of the deformation curves (Fig. 5).

300 Figs. 6(a-d) show the variation of critical strain versus strain rate for AlMg and
301 three AlMgScZr alloys. It can be seen that the relation between the critical strain for
302 the onset of PLC and strain rate is non-monotonic for AlMg and three AlMgScZr
303 alloys. Both the "normal" behavior (the critical strain for the occurrence of PLC
304 increases with the increase in strain rate) and "inverse" behavior (the critical strain for
305 the onset of PLC decreases with the increase in strain rate) are observed for these four
306 materials. Moreover, the type A and A+B/B bands belongs to normal behavior, while
307 the type C band belongs to inverse behavior (Figs. 6(a-d)). The critical strain rate for
308 normal and inverse behaviors, the propagation behaviors of PLC bands as well as the
309 serration amplitude all change with the shearable precipitates becoming nonshearable
310 precipitates, which will be analyzed in details in the following section.

312 Fig. 4 Engineering stress-strain curves of: (a) AlMg (without precipitates), (b)
 313 AlMgScZr-623K/3h (shearable precipitates), (c) AlMgScZr-773K/3h (nonshearable
 314 precipitates), (d) AlMgScZr-813K/24h (nonshearable precipitates). For a better view,
 315 all curves are vertically displaced by 30-50 MPa and the end of the curves is cut. The
 316 upper critical strain rate for the occurrence of PLC is $1.0 \times 10^{-1} \text{ s}^{-1}$ for AlMg and three
 317 AlMgScZr alloys. The lower critical strain rates for the occurrence of PLC are
 318 $2.5 \times 10^{-4} \text{ s}^{-1}$, $1.0 \times 10^{-5} \text{ s}^{-1}$ for AlMgScZr-623K/3h, AlMgScZr-773K/3h alloys. The
 319 lower critical strain rate for the occurrence of PLC is smaller than $1.0 \times 10^{-6} \text{ s}^{-1}$ for
 320 AlMg and AlMgScZr-813K/24 alloys (Due to the limited tensile condition, we cannot
 321 determine the detailed lower values of strain rate for the occurrence of PLC for these
 322 two alloys).

323



324

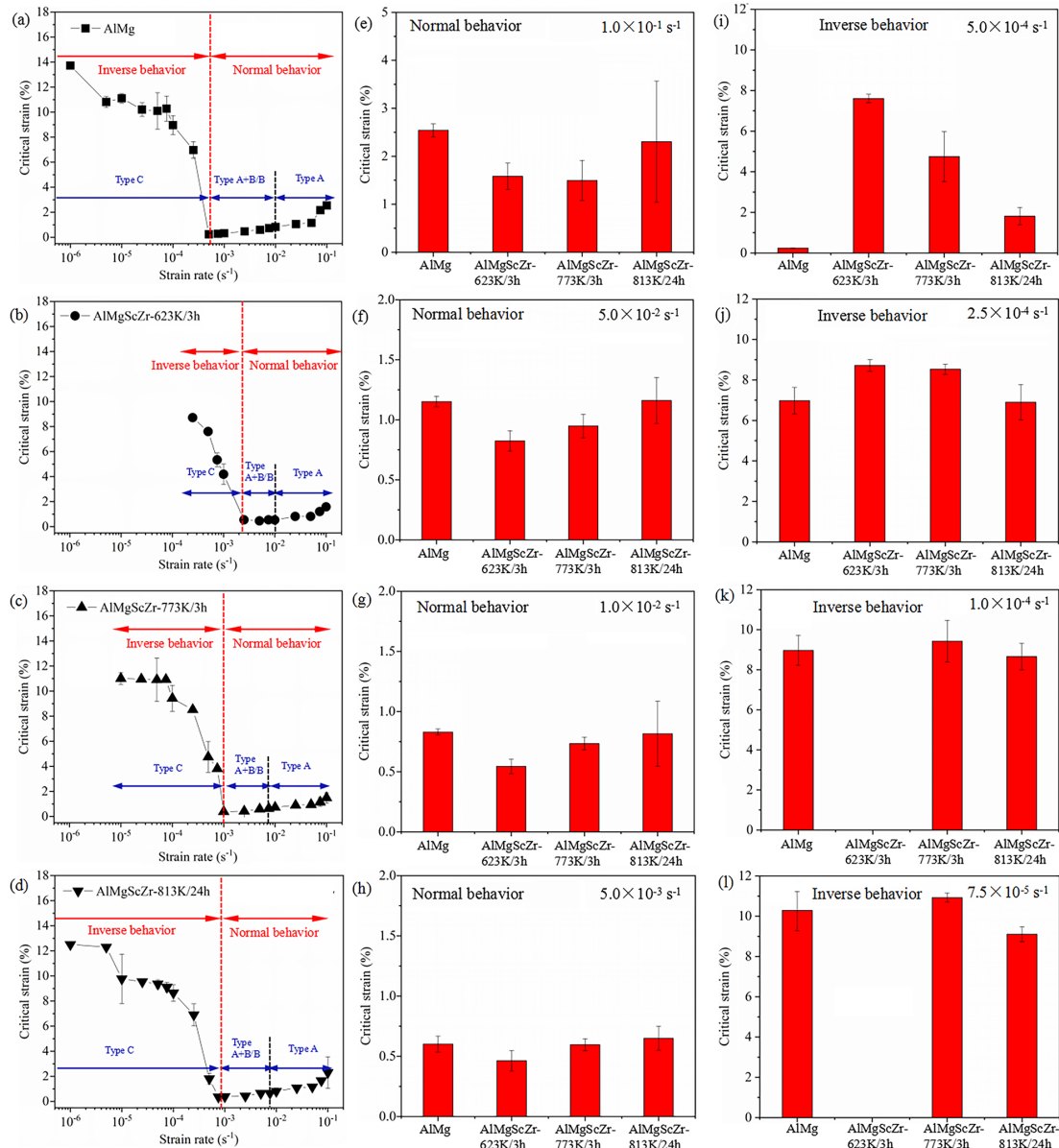
325 Fig. 5 Magnified figure of jerky flow for the AlMg and three AlMgScZr alloys. (a-f)
326 AlMg alloy; (g-l) AlMgScZr-623K/3h alloy; (m-r) AlMgScZr-773K/3h alloy; (s-x)
327 AlMgScZr-813K/24h alloy. For every column, type A (upward steps) first turns to
328 type A+B, then to type B and final to type C (unlocking serrations) with the decrease
329 of strain rate (from top to bottom). The " $\Delta\sigma$ " in the (f) denotes the serration amplitude,
330 which is described in 3.3.3 section.

331

332 **3.3 PLC behavior of AlMg and AlMgScZr alloys**

333 **3.3.1 Influence of the precipitates on onset strain and strain-rate range of PLC**

334 The variations of critical strain and critical strain-rate range of PLC in AlMg and
335 three AlMgScZr alloys is illustrated in Fig. 6 and Table 2. For the AlMg alloy, the
336 strain rate dependence of critical strain exhibits normal behavior in the
337 $5.0 \times 10^{-4} - 1.0 \times 10^{-1} \text{ s}^{-1}$ range and an inverse behavior in the $1.0 \times 10^{-6} - 5.0 \times 10^{-4} \text{ s}^{-1}$ range,
338 as shown in Fig. 6(a) and Table 2. This non-monotonous behavior is also observed for
339 AlMgScZr alloy with precipitates, but the onset strain and strain-rate ranges of PLC
340 change due to the presence of precipitates (Figs. 6(b-d), (e-l) and Table 2). In the
341 normal behavior, shearable precipitates decrease the onset strain of PLC obviously,
342 but this onset strain increases again with the shearable precipitates being replaced by
343 nonshearable precipitates (Figs. 6(e-h)). However, the precipitates have an opposite
344 effect on the onset strain in the inverse behavior regime. That is the shearable
345 precipitates first increase the onset strain significantly, then the onset strain decreases
346 when shearable precipitates become nonshearable precipitates (Figs. 6(i-l)).



347

348 Fig. 6 (a-d) The variation of critical strain versus strain rate for AlMg (without
 349 precipitates), AlMgScZr-623K/3h (shearable precipitates), AlMgScZr-773K/3h
 350 (nonshearable precipitates) and AlMgScZr-813K/24h (nonshearable precipitates). The
 351 black dashed lines are the boundary lines between type A and type A+B/B bands; the
 352 red dashed lines are the boundary lines between type A+B/B and type C bands; the
 353 red dashed lines also are the boundary lines between normal and inverse behaviors;
 354 (e-h) comparison of critical strain for above alloys at different strain rates for normal
 355 behavior: (e) $1.0 \times 10^{-1} \text{ s}^{-1}$, (f) $5.0 \times 10^{-2} \text{ s}^{-1}$, (g) $1.0 \times 10^{-2} \text{ s}^{-1}$, (h) $5.0 \times 10^{-3} \text{ s}^{-1}$; (i-l)
 356 comparison of critical strain for above alloys at different strain rates for inverse
 357 behavior: (i) $5.0 \times 10^{-4} \text{ s}^{-1}$, (j) $2.5 \times 10^{-4} \text{ s}^{-1}$, (k) $1.0 \times 10^{-4} \text{ s}^{-1}$ (PLC disappears for
 358 AlMgScZr-623K/3h alloy), (l) $7.5 \times 10^{-5} \text{ s}^{-1}$ (PLC disappears for AlMgScZr-623K/3h
 359 alloy).

360 Table 2 Strain-rate ranges for normal and inverse behaviors, as well as the critical strain rate
 361 between normal and inverse behaviors for AlMg, AlMgScZr-623K/3h, AlMgScZr-773K/3h and
 362 AlMgScZr-813K/24h alloys.

Samples	Normal behavior range (s ⁻¹)	Inverse behavior range (s ⁻¹)	Critical strain rate (s ⁻¹)
AlMg (without precipitates)	$5.0 \times 10^{-4} - 1.0 \times 10^{-1}$	$1.0 \times 10^{-6} - 5.0 \times 10^{-4}$	5.0×10^{-4}
AlMgScZr-623K/3h (shearable precipitates)	$2.5 \times 10^{-3} - 1.0 \times 10^{-1}$	$2.5 \times 10^{-4} - 2.5 \times 10^{-3}$	2.5×10^{-3}
AlMgScZr-773K/3h (nonshearable precipitates)	$1.0 \times 10^{-3} - 1.0 \times 10^{-1}$	$1.0 \times 10^{-5} - 1.0 \times 10^{-3}$	1.0×10^{-3}
AlMgScZr-813K/24h (nonshearable precipitates)	$7.5 \times 10^{-4} - 1.0 \times 10^{-1}$	$1.0 \times 10^{-6} - 7.5 \times 10^{-4}$	7.5×10^{-4}

363

364 The critical strain rate for which one observes a transition between normal and
 365 inverse behaviors increases evidently when shearable precipitates are introduced.
 366 With increasing the precipitate size (from shearable to nonshearable), this critical
 367 strain rate decreases continuously and comes back to almost the value observed in the
 368 AlMg alloy when the size of nonshearable precipitate is very large (Table 2). On the
 369 other hand, by comparing AlMg and AlMgScZr-623K/3h alloys, the strain-rate range
 370 reduces from $1.0 \times 10^{-1} - 1.0 \times 10^{-6} \text{ s}^{-1}$ to $1.0 \times 10^{-1} - 2.5 \times 10^{-4} \text{ s}^{-1}$ with the introduction of
 371 shearable precipitates (Table 2). However, the strain-rate range expands when
 372 shearable precipitates are replaced by nonshearable precipitates. Finally this
 373 strain-rate range becomes similar to that of the AlMg alloy when the nonshearable
 374 precipitate size is very large (for AlMgScZr-813K/24h alloy). It can be seen that the
 375 influence of precipitates on strain-rate range is mainly focused on the lower critical
 376 strain rate (the lowest strain rate for the appearance of PLC). That is the first increase
 377 and then decrease of lower critical strain rate lead to narrowing and expansion of
 378 strain-rate range with the introduction of shearable precipitates and subsequent
 379 growing into nonshearable precipitates. Therefore, it is concluded that shearable
 380 precipitates inhibit the PLC phenomenon at lower strain rate, but this inhibition effect
 381 is eliminated gradually when shearable precipitates become nonshearable by
 382 coarsening.

383

384 3.3.2 Influence of the precipitates on the propagation behavior of PLC bands

385 It is generally accepted that with the decrease in strain rate type A bands turn into

386 type B and finally form type C bands [Mogucheva et al., 2016; Zhang et al., 2017].
387 The corresponding serrations or steps of type A are periodic, and the serrations are
388 usually locking serrations [Jiang et al., 2007; Hu et al., 2012; Zhemchuzhnikova et al.,
389 2015; Zhang et al., 2017] (Figs. 5(a), (g), (m) and (s)). The locking serrations are
390 characterized by an abrupt rise, followed by a drop to the general level of the
391 deformation curve. Type B bands propagate over short distances discontinuously
392 along the gauge direction and then disappear [Jiang et al., 2007; Hu et al., 2012;
393 Zhemchuzhnikova et al., 2015]. The corresponding serrations are locking and
394 oscillating, as shown in Figs. 5(e), (k), (q) and (w). The serrations corresponding to
395 type C bands are drops in the stress that occur below the general level of the stress
396 strain curve and therefore are considered to be caused by dislocation unlocking
397 [McCormick et al., 1970, 1972b; Jiang et al., 2007; Fu et al., 2012; Zhemchuzhnikova
398 et al., 2015] (Figs. 5(f), (l), (r) and (x)). It should be noted that there is more evident
399 band propagation (smooth intervals) between deep stress drops of type C for
400 AlMgScZr alloys with precipitates [Zhemchuzhnikova et al., 2017] (Figs. 5(r) and
401 (x)). In this work it is found that type A bands appear in the $1.0 \times 10^{-2} - 1.0 \times 10^{-1} \text{ s}^{-1}$
402 range for AlMg alloy (Figs. 5(a) and 6(a)). It should be noted that in this alloy the first
403 emerging PLC band is of type A at the beginning of deformation, and evolves into
404 type A+B bands when the accumulated plastic strain is high, as shown in Fig. 5(b).
405 When strain rate is decreased, mixed type A+B bands dominate the deformation
406 curves, as show in Figs. 5(c) and (d). Type B bands can be superimposed on the type
407 A+B bands and mainly persist at later stages of deformation [Mogucheva et al., 2016]
408 (Figs. 5(d) and (e)). With further decrease of strain rate, type C bands appear for the
409 AlMg sample (Fig. 5(f)). Unlocking type C bands do not occur at the beginning of
410 plastic deformation, and they usually occur when plastic strain reaches a certain
411 amount. The introduction of shearable or nonshearable precipitates does not change
412 the habitual sequence of PLC bands, that is, type A bands still transform to type
413 A+B/B band and finally to type C (Figs. 5(g-l), (m-r) and (s-x)). Precipitation changes
414 only the strain-rate ranges for the appearance of PLC bands, which is depicted in Figs.
415 6(a-d). It can be seen from Fig. 6(a-d) that the type B bands of AlMg sample (in
416 $5.0 \times 10^{-4} - 2.5 \times 10^{-3} \text{ s}^{-1}$ strain-rate range) turn to type C band when shearable
417 precipitates are introduced, and the type B bands of AlMg sample (in
418 $2.5 \times 10^{-3} - 1.0 \times 10^{-2} \text{ s}^{-1}$ strain-rate range) remain. It is demonstrated that the shearable
419 precipitates increase the lower strain rate for the appearance of type C band. With

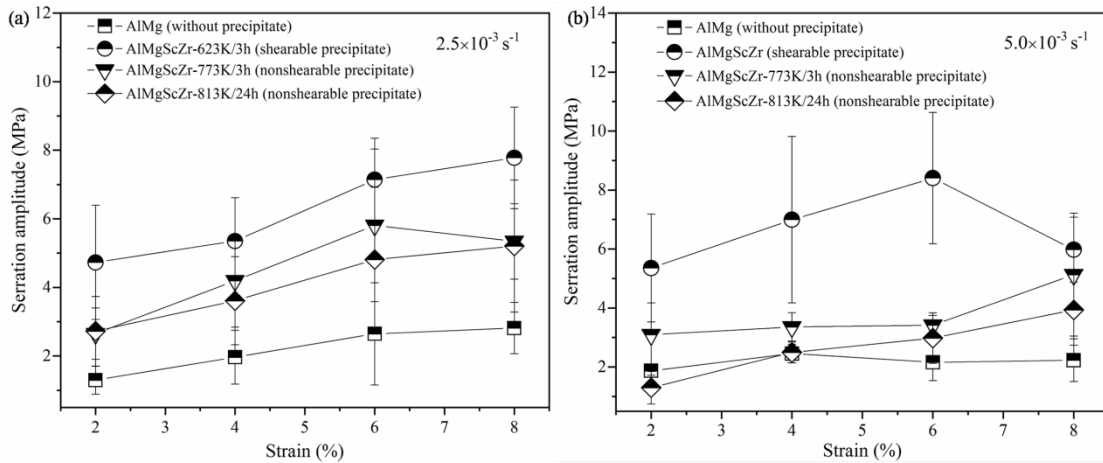
420 shearable precipitates being replaced by nonshearable precipitates, the lower strain
421 rate for the occurrence of type C band decreases, gradually approaching to that of the
422 AlMg alloy when precipitate size is very large (for AlMgScZr-813K/24h alloy).

423

424 **3.3.3 Influence of the precipitates on serration amplitude**

425 The serration amplitude is defined as the height of stress drop, corresponding to
426 the vertical distance between a given local maximum and the latter local minimum on
427 the deformation curve [Pink and Grinberg, 1981; Klose et al., 2004; Zhang et al.,
428 2017], as shown in Fig. 5(f). Fig. 7 shows the variation of the serration amplitude
429 versus strain at normal behavior (at higher strain rate). It is found that the serration
430 amplitude largely increases in case of shearable precipitates and decreases with
431 shearable precipitates becoming nonshearable precipitates (Fig. 7). As the strain range
432 for the appearance PLC is narrow for both AlMg and three AlMgScZr alloys
433 (especially for AlMgScZr-623K/3h, Figs. 4 and 6) in inverse behavior, we do not
434 establish the dependence of serration amplitude versus strain like Fig. 7. We only
435 compare the serration amplitudes of four alloys at a fixed strain, as shown in Fig. 8. It
436 can be seen that the serration amplitude increases with the presence of shearable
437 precipitates, while it decreases when shearable precipitates become nonshearable
438 precipitates. The variation of serration amplitude with the presence of shearable
439 precipitates and subsequent growing into nonshearable at inverse behavior is similar
440 to that at normal behavior.

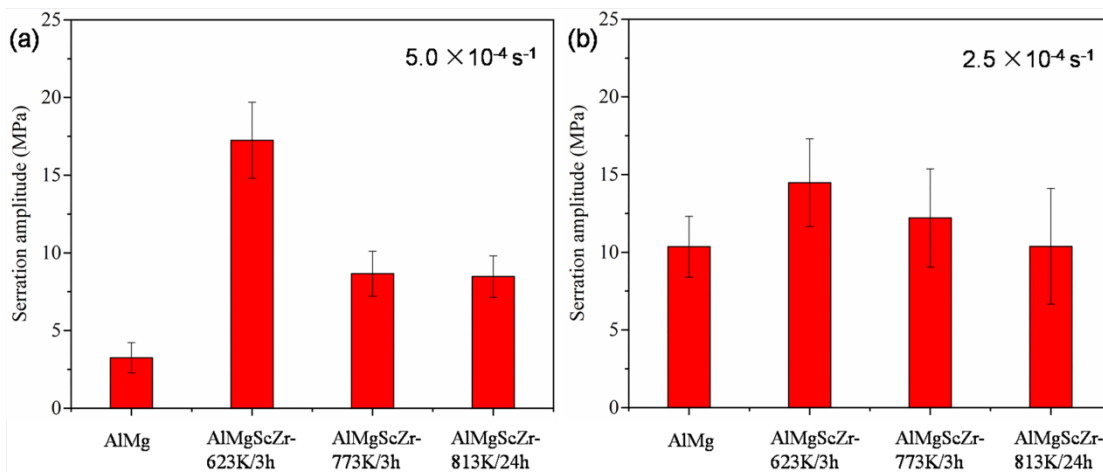
441 No matter for normal behavior or inverse behavior, the magnitude of the serration
442 amplitude manifests a measure for the strength of solutes locking the dislocations
443 [Zhang et al., 2017]. Longer dislocation waiting times are associated with lower
444 nominal strain rates, which allow for more solute diffusion [McCormick, 1972a;
445 Zhang et al., 2017]. Therefore, the waiting time of mobile dislocations influences the
446 serration amplitude. Concerning the above mechanism, it is reasonable to link the
447 different precipitate-dislocation interaction mechanisms to the serration amplitude,
448 which will be discussed in the following.



449

450 Fig. 7 Serration amplitude in AlMg, AlMgScZr-623K/3h, AlMgScZr-773K/3h and
 451 AlMgScZr-813K/24 alloys versus tensile strain at normal behavior (at higher strain
 452 rate): (a) $2.5 \times 10^{-3} \text{ s}^{-1}$; (b) $5.0 \times 10^{-3} \text{ s}^{-1}$. The serrations being analyzed are of type A+B
 453 serration.

454



455

456 Fig. 8 Serration amplitude in AlMg, AlMgScZr-623K/3h, AlMgScZr-773K/3h and
 457 AlMgScZr-813K/24 alloys at same strain (10%) in inverse behavior: (a) $5.0 \times 10^{-4} \text{ s}^{-1}$,
 458 (b) $2.5 \times 10^{-4} \text{ s}^{-1}$. The serrations being analyzed are of type C serration.

459

460 4. Discussion

461 4.1 Strengthening effect induced by $\text{Al}_3(\text{Sc}, \text{Zr})$ precipitates

462 It has been reported that the interaction between dislocation and precipitate changes
 463 from the shearing mechanism to the Orowan bypassing mechanism with the increase
 464 of precipitate size [Ardell, 1985; Gladman, 1999; Knipling et al., 2010; Ma et al.,
 465 2014; Ming et al., 2018]. When precipitate size is small, the dislocations can shear the
 466 precipitate, thus the dominant strengthening mechanism is the shearing mechanism. In
 467 general, the shearing mechanisms mainly include modulus strengthening, coherent

468 strengthening and order strengthening [Ardell, 1985; Knipling et al., 2010; Ma et al.,
469 2014].

470 Firstly, the modulus strengthening, σ_{mod} , due to the difference of shear modulus
471 between precipitate and aluminum matrix, is given by [Ardell et al., 1985; Knipling et
472 al., 2010; Ma et al., 2014; Luca et al., 2019]:

$$473 \quad \sigma_{\text{mod}} = M \cdot 0.0055(\Delta G)^{3/2} (\phi / T)^{1/2} \cdot b(R / b)^{3m/2-1} \quad (1)$$

474 where M is Taylor factor, and $M = 3.06$ for face-centered cubic (FCC) polycrystal
475 materials; ΔG is difference on shear modulus between precipitates G_p ($G_p=78$ GPa for
476 Al_3Sc precipitate) and aluminium matrix G_m ($G_m = 26.2$ GPa for aluminium matrix),
477 thus one has $\Delta G = 41.8$ GPa in this study; ϕ and R are the volume fraction and
478 average radius of precipitates, which can be seen in Table 1; b is the Burgers vector,
479 and $b = 0.286$ nm for aluminum; T is the line tension of dislocation, and $T = 0.5Gb^2 =$
480 1.072×10^{-9} N; m is a constant, and $m = 0.85$.

481 Secondly, the coherent strengthening, σ_{coh} , due to the elastic strain field
482 interactions between coherent precipitates and dislocations, is given by [Ardell et al.,
483 1985; Knipling et al., 2010; Ma et al., 2014; Luca et al., 2019]:

$$484 \quad \sigma_{\text{coh}} = M \cdot \chi (\xi G_m)^{3/2} \cdot (R\phi b / T)^{1/2} \quad (2)$$

485 where χ is a constant, and $\chi = 2.6$; ξ is a misfit parameter, which describes the
486 misfit degree between precipitate and dislocation, and $\xi = 0.0179$.

487 Thirdly, the order strengthening, σ_{ord} , resulting from the formation of antiphase
488 boundaries (APBs), occurs when the mobile dislocations shear the ordered precipitate,
489 which is shown as [Ardell et al., 1985; Knipling et al., 2010; Ma et al., 2014; Luca et
490 al., 2019]:

$$491 \quad \sigma_{\text{ord}} = M \cdot (0.81\eta_{APBs} / 2b) \cdot (3\pi\phi / 8)^{1/2} \quad (3)$$

492 where η_{APBs} is the average APBs energy for (111) plane, and $\eta_{APBs} = 0.5$ J m⁻² for
493 Al_3Sc precipitate.

494 When precipitate size is large, the dislocations cannot shear the precipitate, the
495 dominant strengthening mechanism is Orowan strengthening. The Orowan
496 strengthening σ_{oro} can be expressed as [Knipling et al., 2010; Ma et al., 2014; Luca
497 et al., 2019]:

$$498 \quad \sigma_{\text{oro}} = 0.4MGb \ln(2\bar{R} / b) / \pi L_p \sqrt{1-\nu} \quad (4)$$

499 where \bar{R} is the average planar radius, and $\bar{R} = 0.25R$; L_p is the average
 500 inter-distance of precipitates, and $L_p = R\sqrt{(2\pi/3\phi) - \pi/2}$. The values of L_p for
 501 three AlMgScZr alloys are shown in Table 1.

502 For Al-Sc-Zr based alloy systems, the dominant precipitation strengthening
 503 mechanism depends on the average radius of precipitate [Knipling et al., 2010; Lai et
 504 al., 2013; Luca et al., 2019; Okle et al., 2019]. For smaller precipitates ($R < R_1$)
 505 modulus plus coherency effects are dominant, while above R_1 and below R_2 , order
 506 strengthening becomes important. If the precipitate size overcomes a certain limit
 507 ($R > R_2$), order strengthening is replaced by the strengthening effect associated with
 508 the Orowan bypassing mechanism. The increase in strength as a function of
 509 precipitate size can thus be summarized as follows:

$$510 \quad \Delta\sigma = \begin{cases} \sqrt{\sigma_{\text{mod}}^2 + \sigma_{\text{coh}}^2}, & R < R_1 \\ \sigma_{\text{ord}}, & R_1 < R < R_2 \\ \sigma_{\text{oro}}, & R > R_2 \end{cases} \quad (5)$$

511 Here R_1 is the critical radius between combined modulus-coherent strengthening and
 512 order strengthening; R_2 is the critical radius between order strengthening (or shearing
 513 mechanism) and Orowan strengthening. According to previous studies [Fazeli et al.,
 514 2008; Knipling et al., 2010; Okle et al., 2019; Chen et al., 2021], $R_1 \approx 1.2$ nm and
 515 $R_2 \approx 3-4$ nm for Al-Sc-Zr alloy [Fazeli et al., 2008; Knipling et al., 2010, 2011; Lai et
 516 al., 2013; Luca et al., 2019; Okle et al., 2019]. Besides, we calculate the critical radius
 517 R_c , for the transition between the shearing and bypassing mechanisms, which
 518 according to reference [Bréchet and Louchet, 1990]:

$$519 \quad R_c = \sqrt{3/2}(\Gamma/\gamma) \left[1 + 1.53(\phi/\pi)^{1/2} \right] \quad (6)$$

520 where Γ is the line tension of dislocation; γ is the interface energy. It can be seen
 521 that R_c not only depends on the alloy system (Γ) and precipitate type (γ), but also
 522 depends on the volume fraction of precipitates. Generally, the value of ϕ is much
 523 smaller than π for Al-Sc or Al-Sc-Zr alloys, thus R_c mainly depends on Γ and γ .
 524 According to Eq. (6), the calculated $R_c = 2.8$ nm for Al_3Sc precipitate (for (111) plane,
 525 $\gamma = 0.5 \text{ J m}^{-2}$ [Knipling et al., 2010]) and 3.1 nm (for the (111) plane,

526 $\gamma = 0.445 \text{ J m}^{-2}$ [Lefebvre et al., 2014]) for Al_3Zr precipitate. The calculated values
527 are close to those given by references [Fazeli et al., 2008; Knipling et al., 2010; Lai et
528 al., 2013; Okle et al., 2019], further supporting that the 3 nm is a reasonable value for
529 the critical size between shearing and bypassing mechanisms for Al-Sc-Zr alloy.

530 According to Eq. (5), the dominant strengthening mechanism in
531 AlMgScZr-623K/3h alloy is order strengthening, since the average size is 2.5 nm and
532 belongs to (R_1 , R_2) range. Similarly, the dominant strengthening mechanisms in
533 AlMgScZr-773K/3h and AlMgScZr-813K/24h alloys are Orowan bypassing
534 mechanisms, since their average radius sizes (10.2 and 22.6 nm) are larger than R_2 .

535 In order to verify that the determined strengthening mechanisms are reliable, the
536 results between experiments and strengthening model prediction are compared. Table
537 3 shows the yield stress of AlMg and three AlMgScZr alloys at different strain rates.
538 The estimated yield stress increment as a function of precipitate size is shown in Fig.
539 9 together with the experimental values obtained in this work. For
540 AlMgScZr-623K/3h sample (Fig. 9(a)), the predicted values by combined
541 modulus-coherent strengthening, order strengthening and Orowan strengthening are
542 210, 159, 190 MPa, respectively. The experimental value is about 140 MPa, which is
543 close to the prediction of order strengthening. Thus, it is further verify that the
544 dominant strengthening mechanism in AlMgScZr-623K/3h alloy is the order
545 strengthening. The experimental value is slightly smaller than the prediction of order
546 strengthening, which is due to the overestimated APBs energy of Al_3Sc precipitate.
547 Our results are similar to that of Knipling et al. who reported that the calculated value
548 is higher than the experimental one due to the slightly overestimated APBs energy
549 [Knipling et al., 2010]. For AlMgScZr-773K/3h and AlMgScZr-813K/24h samples
550 (Figs. 9(b) and (c)), the predicted values by combined modulus-coherent
551 strengthening and order strengthening are much higher than experimental values.
552 While the predicted values by Orowan strengthening are very close to the
553 experimental values. Thus, it is further proved that the dominant strengthening
554 mechanisms are Orowan strengthening for these two samples.

555 In general, it can be confirmed that the interaction mechanisms between precipitate
556 and dislocation for three AlMgScZr alloys, providing the theoretical basis for
557 analyzing the influence of precipitates on dislocation motion, distribution and
558 multiplication, and finally providing a good theoretical support for studying the

559 influence of precipitates on the PLC behavior.

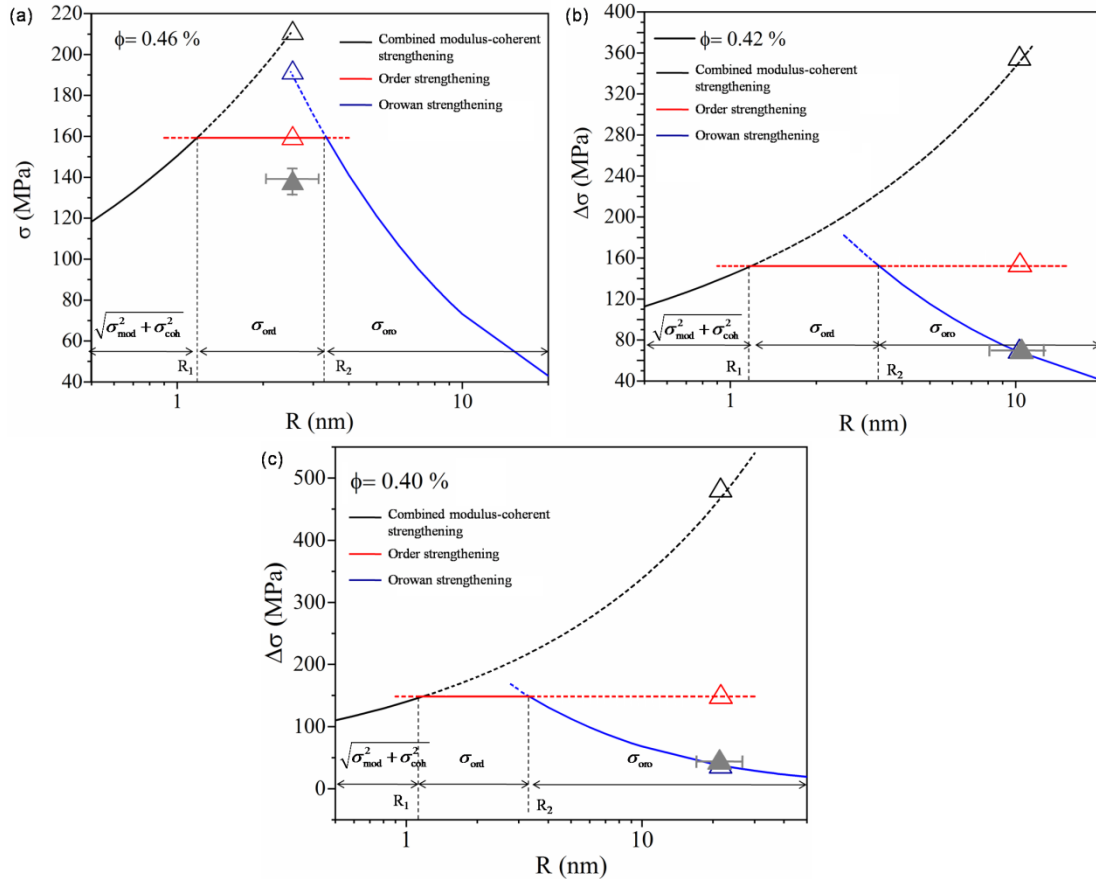
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561 Table 3 Experimentally determined yield stress of AlMg (σ_{s1}), AlMgScZr-623K/3h (σ_{s2}),
 562 AlMgScZr-773K/3h (σ_{s3}) and AlMgScZr-813K/24h (σ_{s4}) alloys.

Strain rate (s ⁻¹)	1×10 ⁻²	5×10 ⁻³	1×10 ⁻³	5×10 ⁻⁴	1×10 ⁻⁴
σ_{s1} (MPa)	168.4±4.8	162.3±1.9	170.7±4.9	173.6±2.9	173.2±4.0
σ_{s2} (MPa)	301.0±4.0	302.0±5.9	305.4±2.5	311.6±16.4	306.0±4.2
σ_{s3} (MPa)	233.8±5.5	235.8±5.5	242.6±19.1	237.2±2.1	235.6±5.1
σ_{s4} (MPa)	216.6±5.5	212.9±10.1	214.8±6.1	221.7±4.9	216.2±2.4

563

564



565

566 Fig. 9 Yield stress increment versus mean precipitate radius for (a)
 567 AlMgScZr-623K/3h [Chen et al., 2021], (b) AlMgScZr-773K/3h [Chen et al., 2021]
 568 and (c) AlMgScZr-813K/24h.

569 of minimum of $\sqrt{\sigma_{mod}^2 + \sigma_{coh}^2}$, σ_{ord} , and σ_{oro} . That is, it is taken the minimum value

570 of $\sqrt{\sigma_{mod}^2 + \sigma_{coh}^2}$, σ_{ord} , and σ_{oro} as the true strengthening increment according to

571 the previous study [Knipling et al., 2010]. The solid symbols in (a-c) represent the
572 yield stress increment obtained from tensile tests. The empty symbols in (a-c) refer to
573 predictions for a certain precipitate radius and volume fraction obtained from SANS:
574 (a) $R = 2.5$ nm, $\phi=0.46\%$; (b) $R = 10.2$ nm, $\phi=0.42\%$; (c) $R = 22.6$ nm, $\phi=0.40\%$.

575

576 **4.2 Effects of precipitates on dislocation motion, distribution and multiplication**

577 The strengthening induced by $\text{Al}_3(\text{Sc}, \text{Zr})$ precipitates is quantified in the 4.1
578 section, proving the different precipitate-dislocation mechanisms for different
579 AlMgScZr samples in this study. According to Fig. 9, order strengthening dominates
580 in the AlMgScZr-623K/3h alloy demonstrating that matrix dislocations shear ordered
581 precipitates and create antiphase boundaries (APBs) on the slip plane in the
582 precipitate. Moreover, Fig. 10 shows the precipitate structure of AlMgScZr-623K/3h
583 alloy after a certain degree of tensile strain (5%). It can be evidently seen that the
584 precipitates are sheared by dislocations in Fig. 10. The APBs energy per unit area can
585 also be interpreted as the force per unit length opposing the dislocation motion when
586 it penetrates the precipitate. During this process, dislocations travel in pairs (so called
587 “superdislocations”), the second dislocation removes the APBs left by the first one
588 [Ardell, 1985]. Recent work of Lefebvre et al. [Lefebvre et al., 2014] revealed that the
589 APBs in Al_3Zr can be the consequence of successive shearing of $[\bar{1}10]/2$ Burgers
590 vector dislocations gliding on distinct (111) planes. Under this circumstance, when the
591 ahead dislocations pass through the precipitate, the subsequent dislocations are prone
592 to move along the specific shear planes owing to its reduced motion resistance,
593 leading to the formation of slip bands (Fig. 11). It can be seen from Fig. 11 that the
594 slip bands with high density of dislocations form. It is clearly shown that lots of
595 dislocation lines parallel with each other in the slip bands, indicating that dislocation
596 pile-ups in slip bands are the results of dislocation motion along above specific shear
597 planes. Although it is difficult to observe the $\text{Al}_3(\text{Sc}, \text{Zr})$ precipitate inside the slip
598 bands, the corresponding SAED of slip bands can confirm the existence of
599 precipitates (Fig. 11(g)). Moreover, we observe the slip bands with relatively low
600 dislocation density with the aid of scanning transmission electron microscope
601 (STEM), as shown in Fig. A1 (supplementary materials). It observes lots of
602 precipitates inside slip bands in the high magnification STEM images. Previous
603 studies also have shown that the slip bands form due to dislocations shearing
604 precipitates [Poole et al., 2005; Xiao et al., 2005; Sharma et al., 2009; Tayon et al.,

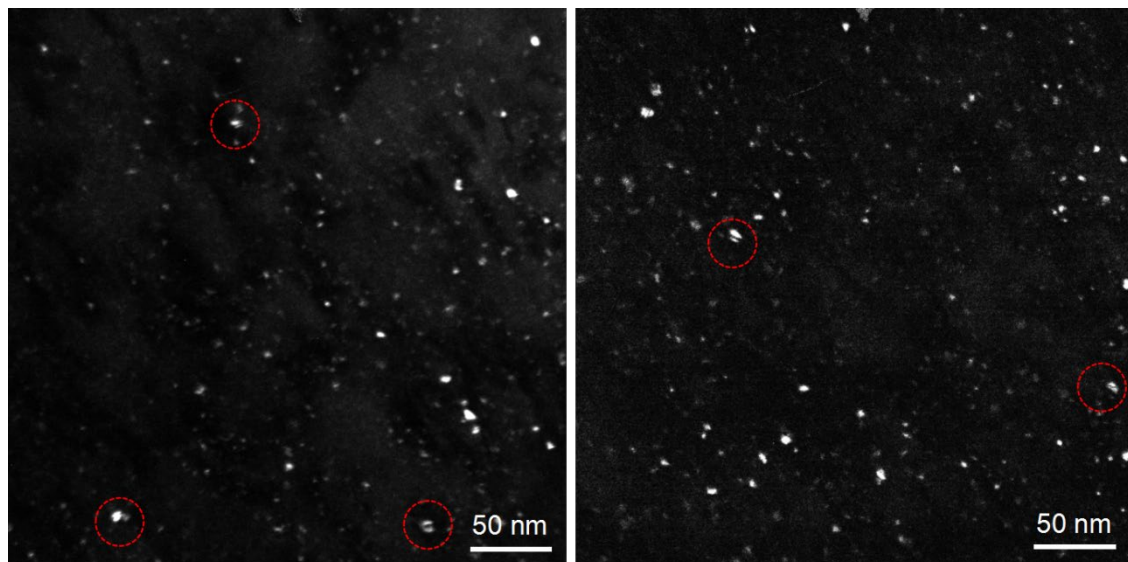
605 [2019; Jaladurgam et al., 2020; Chen et al., 2021](#)], which makes the distribution of
606 dislocations extremely non-uniform. Due to the formation of slip bands, the strain is
607 distributed in a limited volume for the AlMgScZr-623K/3h alloy with shearable
608 precipitates. As a result, the local strain rate and the resultant local strain in the
609 AlMgScZr-623K/3h alloy is higher than the macroscopic equivalent strain. Therefore,
610 although single shearable precipitate cannot accumulate dislocations directly, the
611 shearing process of such precipitates can store a large number of dislocations by the
612 formation of slip bands.

613 With the shearable precipitates becoming nonshearable precipitates, the
614 dislocation shearing mechanism is replaced by dislocation bypassing mechanism (for
615 AlMgScZr-773K/3h and AlMgScZr-813K/24h samples). This process leaves Orowan
616 loops around the nonshearable precipitates. The dislocations are distributed relative
617 homogeneously during tensile deformation, which is different from that of
618 AlMgScZr-623K/3h alloy with shearable precipitates, as shown in Figs. 12(b) and (c).
619 It should be noted that the dislocations are also distributed uniformly for the AlMg
620 sample without precipitates (Fig. 12(a)). Our results are similar to Poole et al. who
621 reported that the shearable and nonshearable precipitates had distinct effects on
622 dislocation distribution through different precipitate-dislocation interactions [[Poole et
623 al., 2005](#)].

624 In order to quantitatively evaluate the dislocation density caused by precipitates,
625 the dislocation densities of AlMg, AlMgScZr-623K/3h, AlMgScZr-773K/3h and
626 AlMgScZr-813K/24h were evaluated by XLPD (the detailed evaluation procedure can
627 be found in reference [[Borbély and Groma, 2001](#)]). The corresponding (200) peak
628 profiles (as an example) and variations of dislocation densities versus tensile strain
629 are shown in Fig. 13. The dislocation density of the AlMgScZr-623K/3h sample is
630 larger by one order of magnitude compared with that of the AlMg alloy. Dislocation
631 density decreases when shearable precipitates become nonshearable precipitates. The
632 results are in very good agreement with the TEM images (Figs. 11 and 12). It should
633 be noted that the AlMgScZr-623K/3h alloy has the highest dislocation density in all
634 studied alloys, which is mainly due to its higher initial dislocation density retained
635 during hot extrusion. Although the increment on dislocation density in

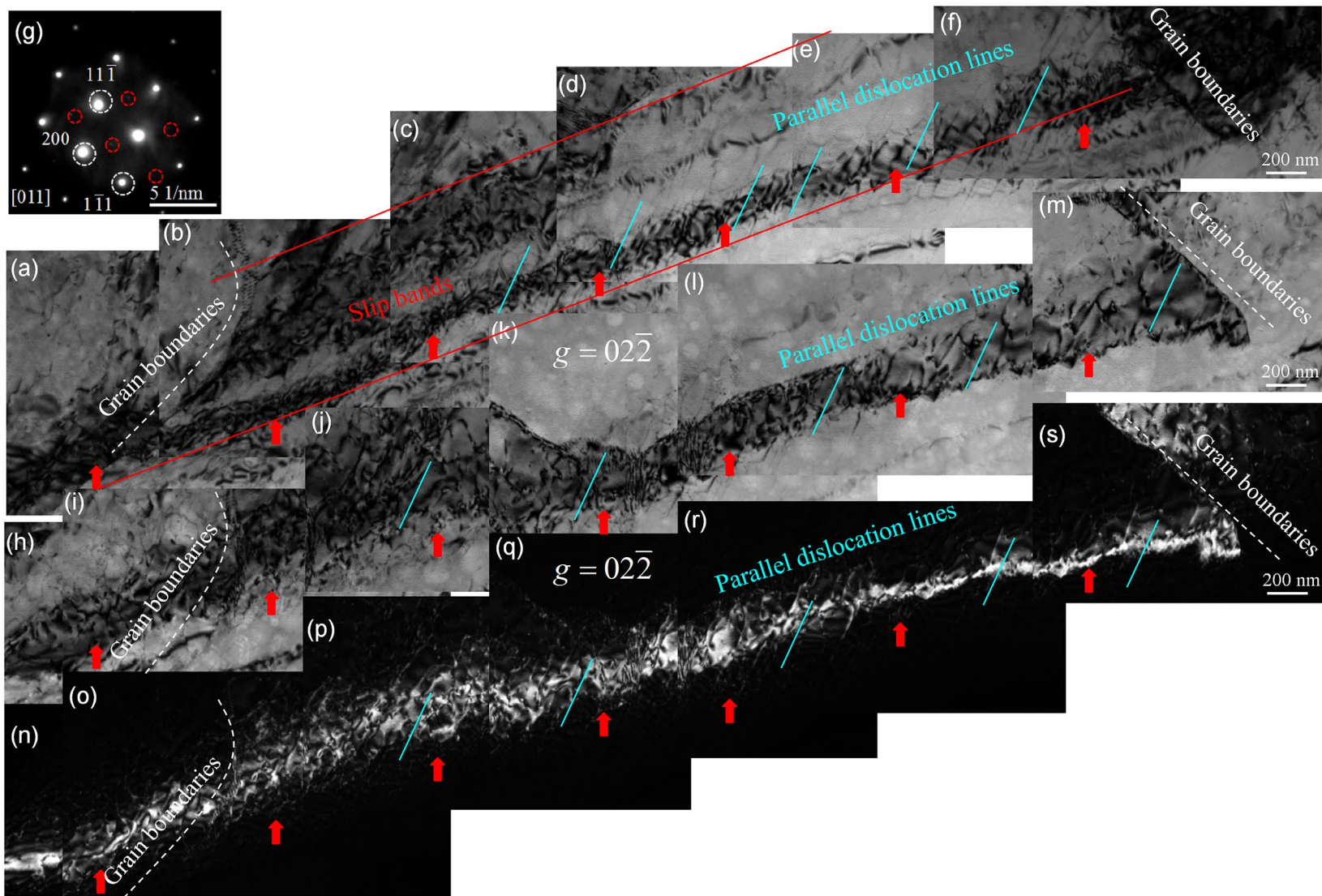
636 AlMgScZr-623K/3h alloy at the beginning stage of tensile deformation is higher than
637 the others, the differences on dislocation density increment caused by tensile
638 deformation for these four samples are relatively small. The AlMgScZr-623K/3h alloy
639 with shearable precipitates has a higher dislocation density storage after hot extrusion
640 deformation, which is mainly due to the fact that the shearable precipitates have
641 stronger pinning effect on dislocations [Humphreys and Hatherly, 2004; Fazeli et al.,
642 2008]. Thus, more dislocations can be retained during hot extrusion. The dislocations
643 in AlMg alloy without precipitates cannot be arrested and pinned effectively by
644 precipitates, thus the AlMg alloy has the lowest dislocation density storage after hot
645 extrusion. Compared with AlMg alloy, the AlMgScZr-773K/3h and
646 AlMgScZr-813K/24 alloys have a higher density of dislocation storage due to pinning
647 effect by the nonshearable precipitates, but their stored dislocation densities are
648 smaller than that of AlMgScZr-623K/3h alloy, for their pinning effect on dislocations
649 is weaker than that of shearable precipitates [Fazeli et al., 2008].

650

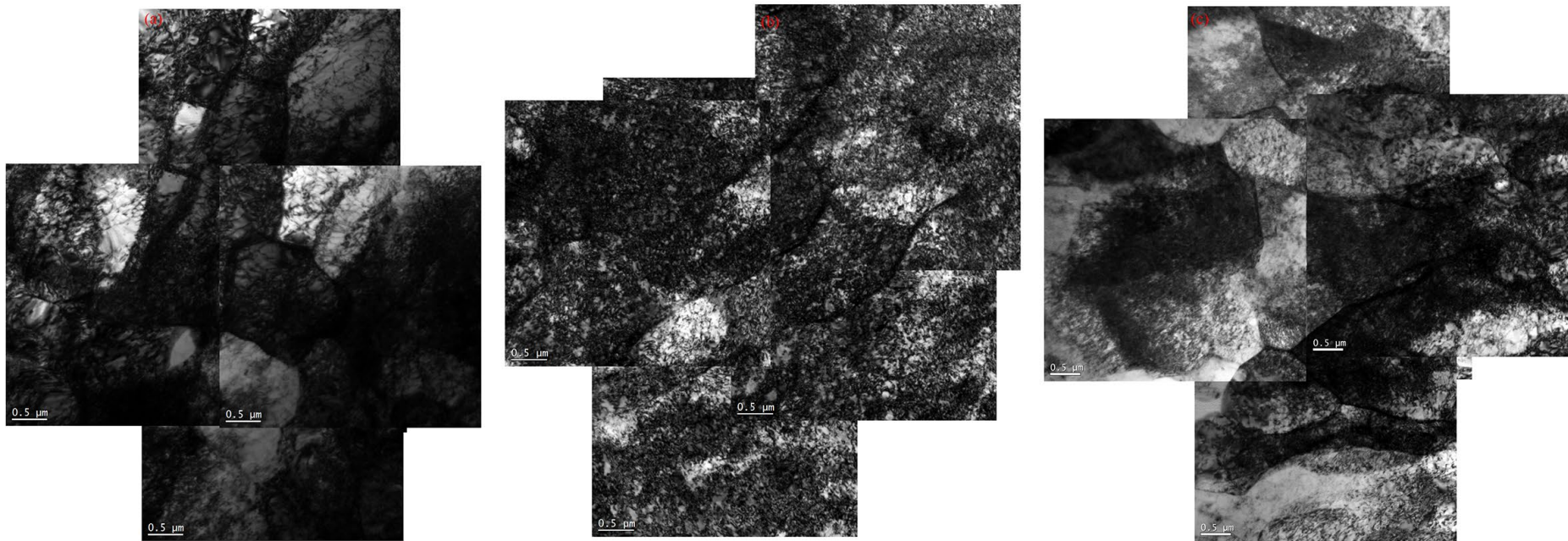


651

652 Fig. 10 Dark-field TEM showing the $\text{Al}_3(\text{Sc}, \text{Zr})$ precipitates for AlMgScZr-623K/3h
653 alloy after 5% tensile strain. The zone axis is [011]. The dashed red circles clearly
654 show that the precipitates are sheared by dislocations.



656 Fig. 11 TEM images of dislocation structure for AlMgScZr-623K/3h alloy with shearable precipitates after 5% tensile strain: (a-f) bright TEM
 657 images of slip bands with [011] zone axis; (g) is the corresponding SAED of (a-f); (h-m) bright TEM images of slip bands with dual beam
 658 diffraction ($g=02\bar{2}$); (n-s) corresponding dark TEM images of (h-m) with dual beam diffraction ($g=02\bar{2}$). Here the red full lines in (a-f) refer to
 659 the slip bands due to dislocation shearing precipitates. The slip bands labelled by red arrows in the (a-f), (g-l) and (m-s) are the same slip bands.
 660 The cyan dotted lines refer to the parallel dislocation lines, and the dashed white lines denote grain boundaries.
 661



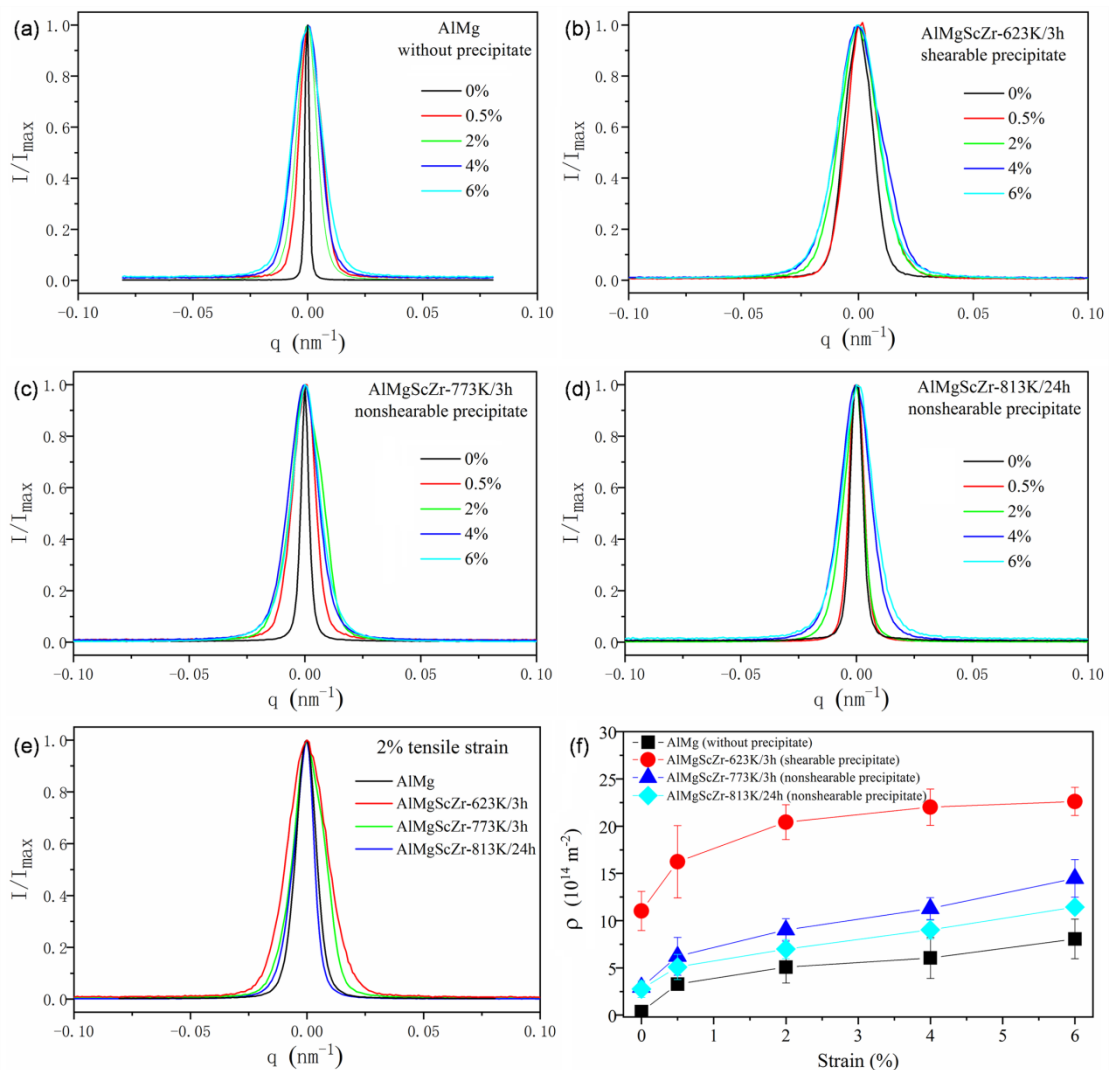
662
 663 Fig. 12 Bright TEM images of dislocation structure for (a) AlMg (without precipitates), (b) AlMgScZr-773K/3h (with nonshearable precipitates)
 664 and (c) AlMgScZr-813K/24h (with nonshearable precipitates) alloys after 5% tensile strain. The TEM images of these three alloys are acquired
 665 with the [011] zone axis. The dislocations of above three alloys distribute relative uniformly compared with AlMgScZr-623K/3h alloy.

666 The TEM images of all samples after 5% tensile strain are presented to
667 demonstrate the magnitude of dislocation density and dislocation distribution after
668 tensile deformation (Figs. 11 and 12). For AlMgScZr-623K/3h alloy, after tensile
669 deformation, lots of slip bands is observed in microstructure, and dislocations pile up
670 inside these slip bands. It is reasonable to infer that the retained dislocations during
671 hot extrusion can move along a specific direction to form slip bands due to the
672 shearing mechanism between precipitates and dislocations (this process may also
673 contain lots of new dislocations generated during tension). On the other hand, once
674 the dislocations move and form slip bands, these dislocations are difficult to be
675 annihilated during tension. This is because most of these dislocations inside slip bands
676 have the same Burgers vector (most of dislocations lines are parallel, indicating that
677 these dislocations are of same Burgers vector), thus they are difficult to be annihilated
678 through cross-slip [Sandström, 1977; Nes, 1995; Xiao et al., 2005]. It should be noted
679 that there are still a considerable number of dislocations outside the slip bands in
680 AlMgScZr-623K/3h alloy, although its local density is much lower than that in slip
681 bands. The measured dislocation density by SRXRD is total dislocation density,
682 which contains the dislocations both inside slip bands and outside the slip bands. For
683 AlMg and other AlMgScZr alloys, the dislocations distribute more uniformly (Fig.
684 12), these tangled dislocations usually have different Burgers vectors. Dislocation
685 annihilation is easy to occur when dislocation tangles become complicated, which
686 also limits the rapid increase of dislocation density [Verdier et al., 1998; Kocks and
687 Mecking, 2003]. Thus, although alloys with nonshearable precipitates can accumulate
688 lots of Orowan loops during tension, the total density of dislocations for a given
689 microscopic tensile strain is much smaller (Fig. 13(f)) than in the case of shearable
690 precipitates. The AlMgScZr-623K/3h alloy has the higher dislocation density than
691 AlMg and the other AlMgScZr alloys is more related to the formation of slip bands
692 (dislocations can be stored more effectively inside the local slip bands).

693 The analysis in Figs. 11, 12 and 13 demonstrates that shearable and nonshearable

694 precipitates have different effects on dislocation storage during tensile deformation by
 695 changing dislocation motion and distribution. The change of dislocation distribution
 696 and density of stored dislocation for a given macroscopic tensile strain will affect the
 697 waiting time of dislocation in the localized bands and diffusion of solute atoms during
 698 DSA, thus inevitably influencing the propagating behavior, serration amplitude, onset
 699 strain and strain-rate range of PLC, which are discussed as follows.

700



701

702 Fig. 13 Peaks profiles of AlMg and three AlMgScZr alloys as well as the quantitative
 703 measurement of dislocation density: (200) peaks profiles of (a) AlMg, (b)
 704 AlMgScZr-623K/3h, (c) AlMgScZr-773K/3h, and (d) AlMgScZr-813K/24h after
 705 different tensile strains; (e) (200) peaks profiles of AlMg and three AlMgScZr alloys
 706 after 2% tensile strain; (f) variations of dislocation density versus tensile strain for

707 AlMg and three AlMgScZr alloys. The dislocation density is measured averagely
708 from (200), (220) and (311) peaks. Here the peaks profiles of (220) and (311) are
709 omitted for simplicity, since they have similar features to the (200) peaks.

710

711 **4.3 Influence of precipitates on the characteristics of PLC bands**

712 The PLC is due to the interaction between solute atoms and dislocations, which is
713 the process dominated by dislocation pinning and unpinning [Kubin and Estrin, 1985;
714 Rizzi and Hähner, 2004]. However, the solute-dislocation interaction mechanisms for
715 the appearance of PLC are different in normal and inverse behavior regimes
716 [Chaenock, 1969; Soler-gomez and Tegart, 1969; McCormick et al., 1970, 1972b; Fu
717 et al., 2012; Zhemchuzhnikova et al., 2015; Zhang et al., 2017; Geng et al., 2020]. In
718 normal behavior regime (at higher strain rate), the mobile dislocations usually have a
719 higher movement velocity and the solute atoms cannot capture the mobile dislocations
720 due to their relatively lower diffusion rate. When the mobile dislocations are arrested
721 by forest dislocations and other obstacles, the solute atoms have the possibility to
722 capture the mobile dislocations. Only when the solute atoms diffuse to the arrested
723 mobile dislocations during the waiting time does PLC occur. Thus, at normal behavior,
724 the PLC appears in the condition of the solute atoms diffusing to the arrested mobile
725 dislocations and pinning these dislocations. The stress strain curves usually exhibit
726 locking serrations or steps (Figs. 5(a), (c)), and the PLC appears in the form of an
727 abrupt rise. The abrupt rise indicates a process that the solute atoms pin the arrested
728 mobile dislocations, also meaning that the solute atoms pinning the mobile
729 dislocations leads to PLC. Therefore, at this time, the physical mechanism for
730 controlling the appearance of PLC is defined as “dislocation pinning mechanism”. On
731 the contrary, in inverse behavior regime (at lower strain rate), the mobile dislocations
732 can be pinned by solute atom atmosphere before the tensile plastic deformation due to
733 the very low velocity of mobile dislocations. The critical condition for the appearance
734 of PLC is the breakaway of mobile dislocations from this atmosphere, which is
735 different from that in normal behavior regime [Chaenock, 1969; Soler-gomez and
736 Tegart, 1969; McCormick et al., 1970, 1972b; Fu et al., 2012; Geng et al., 2020]. The
737 stress strain curves usually manifest unlocking serrations (Fig. 5(f)) for inverse
738 behavior, and the PLC appears in the form of a stress drop. The stress drop indicates a
739 process that the mobile dislocations escape from the pinning of solute atmosphere,

740 meaning that the escape of mobile dislocations from the existing atmosphere also
741 leads to the PLC. The physical mechanism controlling the appearance of PLC is thus
742 defined as “dislocation unpinning mechanism”.

743 It is shown that precipitates influence the transition between A+B/B and C type
744 bands: the corresponding strain rate firstly increases from $5.0 \times 10^{-4} \text{ s}^{-1}$ to $2.5 \times 10^{-3} \text{ s}^{-1}$
745 for the AlMgScZr-623K/3h alloy with shearable precipitates followed by a decrease
746 from $2.5 \times 10^{-3} \text{ s}^{-1}$ to $7.5 \times 10^{-4} \text{ s}^{-1}$ when shearable precipitates become nonshearable
747 precipitates (Figs. 6(a-d)). These results show that precipitates increase the lower
748 strain rate for the occurrence of type C band. Our results are similar to Cai et al. who
749 manifested that precipitates increase the lower strain rate for the appearance of type C
750 band in Ni-based superalloys [Cai et al., 2015, 2017]. Regarding as sites where the
751 dislocations pile up (Fig. 11), shearable precipitates increase the internal stress and
752 favor a change from type B band to type C band with the decrease of strain rate (Figs.
753 6(a-d)). The shearable precipitates strongly influence the dynamical mechanisms of
754 local strain propagation, through the localization of dislocation glide (Fig. 11)
755 [Mogucheva et al., 2016].

756 The type C bands are a kind of dislocation unlocking serrations due to the
757 unpinning mechanism between mobile dislocation and solute atoms, while type A and
758 B bands are caused by dislocation pinning mechanism. Shearable precipitates increase
759 the lower strain rate for the formation of type C band (Fig. 6(a-d)), meaning that the
760 shearable precipitates increase the critical strain rate between dislocation unpinning
761 and pinning mechanisms. It is because that, at same strain rate, the high local
762 dislocation density in alloy containing shearable precipitates could enhance the
763 diffusion rate of solute atoms (Fig. 13(f)). On one hand, the higher density of
764 dislocation can reduce the diffusion activation energy and thus enhances the diffusion
765 velocity [Legros et al., 2008]. On the other hand, the substitutional solute migration is
766 correlated with the vacancy concentration [Picu and Zhang, 2004; Picu and Xu, 2007;
767 Van der Ven et al., 2010]. As such systems with larger dislocation content are more
768 likely to exhibit local vacancy supersaturation which will aid diffusion [Van der Ven
769 et al., 2010]. When the diffusion rate of solute atoms increases, it is much easier to
770 meet the condition of dislocation unpinning (i.e. the diffusion rate of solute atoms is
771 much higher than the velocity of mobile dislocations). At this time, there is no need to
772 decrease the movement velocity of mobile dislocation to meet dislocation unpinning
773 via decreasing the strain rate. Therefore, the shearable precipitates increase the critical

774 strain rate between unpinning and pinning mechanisms by increasing local dislocation
775 density for a given microscopic strain. However, these effects (increasing the lower
776 strain rate for the formation of type C band and increasing the critical strain rate
777 between unpinning and pinning mechanisms) weakens due to the decrease of stored
778 dislocation density when shearable precipitates become nonshearable precipitates
779 (Figs. 6(a-d) and Table 2).

780 As far as the serration amplitude is concerned, the appearance of precipitates
781 increases the magnitude of serration amplitude no matter for normal behavior (Fig. 7)
782 or inverse behavior (Fig. 8). The serration amplitude $\Delta\sigma$ can be expressed as [Chen
783 et al., 1996; Zhang et al., 2017]:

$$784 \Delta\sigma \propto \left[\dot{\varepsilon}^{-1} \varepsilon^{\beta(1/2+\gamma)} d^{-n(1/2+\gamma)} T^{-1} \exp(-Q/kT) \right]^{2/3} \quad (7)$$

785 where $\dot{\varepsilon}$ is strain rate; ε is strain; d is the grain size; T is temperature; Q is the
786 activation energy for solute diffusion, which is associated with solute-vacancy site
787 exchange; k is the Boltzmann constant. Precipitates enhance the accumulation of
788 dislocation density, which can reduce Q and thus accelerate the diffusion of atoms
789 [Legros et al., 2008]. Therefore, it is reasonable to infer that the $\Delta\sigma$ increases for
790 AlMgScZr alloy with precipitates in contrast to AlMg alloy based on Eq. (7). By
791 comparing the samples with shearable and nonshearable precipitates, the shearable
792 precipitates can reduce Q obviously by storing more dislocations, leading to a larger
793 serration amplitude.

794

795 **4.4 Influence of precipitates on normal and inverse behaviors**

796 Since the solute-dislocation interactions for the appearance of PLC at normal and
797 inverse behaviors are related to the dislocation pinning and unpinning mechanisms,
798 respectively [Chaenock, 1969; Soler-gomez and Tegart, 1969; McCormick et al.,
799 1970, 1972b; Fu et al., 2012; Geng et al., 2020]. The introduction of shearable
800 precipitates and nonshearable precipitates have different effects on different
801 solute-dislocation interactions, which are discussed as follows.

802

803 **4.4.1 Influence of precipitates on normal behavior**

804 At higher strain rates, the locking serration (type A and B bands) or upward step

805 (type A bands) demonstrates that mobile dislocations are not pinned by diffusing
806 solute atoms at the beginning of tensile deformation [Chaenock, 1969; Soler-gomez
807 and Tegart, 1969; McCormick et al., 1970, 1972b; Fu et al., 2012; Geng et al., 2020].
808 When the mobile dislocations are captured by diffusing solute atoms, the PLC occurs.
809 For the AlMg alloy mobile dislocations are mainly pinned by forest dislocations,
810 while in AlMgScZr the mobile dislocations are pinned both by forest dislocations and
811 precipitates together. When the shearable precipitates are present, the mobile
812 dislocations will be additionally pinned until they can find a partner dislocation to
813 form the superdislocations and cut through the precipitates. Thus, the pinning effect
814 on mobile dislocations increases evidently when the shearable precipitates appear.
815 As a consequence, the waiting time t_w also increases obviously. When shearable
816 precipitates become nonshearable, the mobile dislocations will be additionally pinned
817 until they pass through the precipitates. Compared with the shearable precipitates, the
818 pinning effect on mobile dislocations provided by nonshearable precipitates weakens
819 (Fig. 9 and Table 3). Thus, the waiting time t_w decreases with the shearable
820 precipitates transforming to nonshearable. Here t_w refers to the time during which the
821 mobile dislocations are arrested by obstacles, which is given by [Bréchet and Estrin,
822 1995; Curtin et al., 2006]:

$$823 \quad t_w = \Omega / \dot{\epsilon} = \rho_m b L / \dot{\epsilon} \quad (8)$$

824 where Ω is elementary plastic strain [Bréchet and Estrin, 1995]; ρ_m is denoted as
825 mobile dislocations; b is Burgers vector; L is the average spacing between obstacles,
826 representing the flight distance of solute atom migration; $\dot{\epsilon}$ is the strain rate.
827 Compared to the AlMg alloy, the waiting time t_w of mobile dislocations in
828 AlMgScZr alloy with shearable precipitates increases, which indicates that the solute
829 atoms have more time to capture the arrested mobile dislocations. At this time, there
830 is no need to enhance the diffusion velocity of solute atoms by imposing a higher
831 tensile strain, and thus the onset strain of PLC decreases. When the shearable
832 precipitates transform into nonshearable precipitates, the waiting time t_w decreases,
833 and the solute atoms need a higher diffusion velocity in order to capture the arrested
834 mobile dislocations. At this time, a relative higher tensile strain is required for
835 generating more vacancies to enhance diffusion velocity of solute atoms, thus the
836 onset strain increases. In summary, the presence of shearable precipitates and their

837 subsequent transforming into nonshearable precipitates make the imposed onset strain
838 to meet the pinning requirement first decreased and then increased at high strain rates,
839 as shown in Figs. 6(e-h).

840

841 **4.4.2 Influence of precipitates on inverse behavior**

842 At lower strain rates, the unlocking serration (type C band) indicates that mobile
843 dislocations are pinned before tensile plastic deformation. When the mobile
844 dislocations escape from pinning obstacles, the PLC occurs [Chaenock, 1969;
845 Soler-gomez and Tegart, 1969; McCormick et al., 1970, 1972b; Fu et al., 2012; Geng
846 et al., 2020]. Thus the PLC phenomenon will occur after a certain degree of tensile
847 plastic deformation. In the process of mobile dislocations overcoming obstacles, the
848 mobile dislocations need to overcome the energy barrier. It is obvious that the energy
849 barrier will increase due to the introduction of precipitates and the strain rate for the
850 onset of PLC will increase correspondingly. The results in sections 4.1 and 4.2 have
851 proved that the shearable precipitates improve the strength and store a higher density
852 of dislocation density, indicating that the applied stress will be increased if the mobile
853 dislocations expect to overcome the energy barriers. However, the inhibition effects
854 decrease due to the shearable precipitates becoming nonshearable precipitates, and as
855 a result the energy barrier will be decreased due to the weaker precipitate-dislocation
856 interaction (the subsequent decrease of strength and dislocation density can prove it
857 (Figs. 9 and 13)). Hence, the shearable precipitates will make the low critical strain
858 rate for the occurrence of PLC increased evidently, and this low critical strain rate will
859 decrease with the shearable precipitates being replaced by nonshearable precipitates.
860 This explains that the shearable precipitates firstly narrow the strain-rate ranges for
861 the appearance of PLC and then expands this strain-rate range with the shearable
862 precipitates becoming nonshearable (Fig. 6(a-d) and Table 2). For the onset strain, if
863 the strain rate is same, the first increase and then decrease of energy barriers by
864 introducing shearable precipitates and subsequent growing into nonshearable
865 precipitates result in the corresponding variations of onset strain. It is because
866 dislocations can escape the higher energy barrier only by increasing onset strain to
867 provide a higher stress. Therefore, at inverse behavior regime, the shearable
868 precipitates first enlarge the critical strain for the appearance of PLC and then
869 decreases this strain when shearable precipitates become nonshearable precipitates

870 (Figs. 6(i-l)).

871 It should be noted that the process of dislocation shearing small precipitates is
872 influenced by thermal activation for AlMgScZr-623K/3h alloy according to Argon's
873 study [Argon, 2008]. At higher strain rate, the mobile dislocations have a higher
874 movement velocity, which is easier to cutting the precipitates and overcome the
875 energy barrier, since the mobile dislocations have a higher kinetic energy. Thus, at this
876 time, the dislocations have a higher cutting rate, and the waiting time provided by
877 precipitates will decrease to some extent.

878

879 **5. Conclusions**

880 In this study, we systematically investigate the influence of shearable and
881 nonshearable Al₃(Sc, Zr) precipitates on PLC behavior for precipitation hardening
882 AlMgScZr alloys. It is shown that the presence of precipitates, as well as the change
883 of the precipitate-dislocation interaction mechanism leads to different PLC behavior.
884 The main conclusions are the following:

885

886 (1) Shearable precipitates promote the formation of slip bands with dislocation
887 pile-ups due to their precipitate-dislocation interaction, while the dislocations
888 distribute relatively uniformly in samples containing nonshearable precipitates.
889 Compared with the samples with nonshearable precipitates, the higher local strain rate
890 and higher local strain caused by shearable precipitates make more dislocations
891 retained in samples, leading to a higher dislocation storage.

892

893 (2) Shearable precipitates increase the lower strain rate for the appearance of type C
894 band, but this effect decreases with the shearable precipitates becoming nonshearable
895 precipitates. Shearable precipitates enhance the serration amplitude largely, but the
896 serration amplitude decreases when shearable precipitates become nonshearable
897 precipitates. Both above phenomena are attributed to the first increase and then
898 decrease of dislocation density controlled by different dislocation-precipitate
899 interactions.

900

901 (3) At higher strain rates, the shearable precipitates decrease the onset strain but the
902 onset strain then increases when shearable precipitates become nonshearable

903 precipitates, which is due to the first increase and then decrease of waiting time
904 controlled by different precipitates. At lower strain rates, the shearable precipitates
905 increase the onset strain but the onset strain then decreases with the shearable
906 precipitates being replaced by nonshearable precipitates, which is due to the first
907 increase and then decrease of energy barrier which needs to be overcome.

908

909 (4) Shearable precipitates narrow the strain-rate range of PLC, but this strain-rate
910 range expands again when shearable precipitates are replaced by nonshearable
911 precipitates. That is because the lower critical strain rate being dominated by the
912 precipitate-induced energy barrier first increases and then decreases.

913

914 (5) At higher strain rates, precipitates act as additional obstacles, along with forest
915 dislocations, to arrest mobile dislocations. Solute atoms thus have more time to
916 diffuse onto arrested mobile dislocations, leading to a decrease of onset strain, i.e.,
917 precipitates facilitate dislocation pinning process. At lower strain rate, additional
918 pinning effect provided by precipitates results in a higher required applied stress for
919 arrested mobile dislocations to escape from the pinning of atmosphere, leading to an
920 increase of onset strain, i.e., precipitates prevent dislocation unpinning process. Both
921 of the above effects attenuate with the shearable precipitates becoming nonshearable
922 precipitates.

923

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