Analysis on the hot deformation flow stress curves of novel 6082 aluminium alloys with Mn addition

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Abstract. Two types of 6082 alloy, base alloy free of Mn and novel 0.75Mn alloy with an addition of 0.75% Mn, were prepared. A heat treatment of 450 °C for 6 h was performed, during which large numbers of α-Al (Fe, Mn) Si dispersoids precipitated in 0.75Mn alloy. Uniaxial compression tests were carried out and a series of flow stress curves were obtained at temperature range of 400–550 °C and at strain rate range of 0.001s^-1–1s^-1. The results revealed an impressive difference between the flow stress curves of base and 0.75Mn alloy that after the yield plateau, the flow stresses of base alloy remained fairly steady at certain levels while the flow stress in 0.75Mn alloy decreased along with strain until the end of deformation. The reason for the decrease of flow stress in 0.75Mn alloy was neither the dissolution of dispersoids nor the dynamic recrystallization. The decrease of flow stress was attributed to the loss of Mn solid solution strengthening effect based on the Mn diffusion rate-controlling mechanism and “pipe diffusion”.

1. Introduction

The main strengthening mechanism of traditional 6082 aluminium alloys at room temperatures is based on the fine nano-scale Mg²Si precursors [1], [2]. However, the coarsening and/or dissolution of Mg²Si starts to take place at the temperature exceeding 200 °C (similar as over aging), which leads to severe deterioration of mechanical properties [3].

To develop novel alloys that can be applied at elevated temperatures, our previous study explored the way using α-Al (Mn, Fe) Si dispersoids to strengthen 6082 alloy at elevated temperatures [4]. The results show that large number of dispersoids was induced by adding Mn element and through a low-temperature homogenization of 450 °C for 6 h. The dispersoids-strengthened alloy yielded significantly increased high-temperature flow stresses. An increase in hot deformation activation energy from 191 kJ/mol for the Mn-free alloy to 301 kJ/mol for the alloy with 0.75% of Mn was derived from a series of hot deformation flow stress curves.

The feature of flow stress curves are important indicators to the changes in microstructures and can reflect the balance between dynamic hardening and dynamic softening during the hot deformation process [5], [6]. In the typical hot deformation flow stress curves, the flow stress usually increases sharply at the early stage of deformation until reaching a peak plateau [7], [8]. After then, flow stress curves features can be generally divided into several typical cases: i) increase, ii) stay stable and iii) decrease along with increasing strain. The different trends of flow stresses after the peak plateau are closely related the hot deformation conditions [9], [10] and microstructural evolutions.
For the first case of the continuous increase of flow stress along with strain, it is usually associated with the condition when the work hardening overcome the effect of dynamic softening. Shakiba [11] studied the flow stress behavior of a 1xxx dilute aluminium alloy, the resulted shows that continuous increased flow stresses after yield plateau were observed at the low temperature and high strain rate deformation temperatures of 350°C/1s\(^{-1}\) when dynamic recovery (DRV) was the predominant softening mechanism. While with the increased deformation temperature by 400°C at the fixed strain rate of 1s\(^{-1}\) or at a lowered strain rate of 0.001s\(^{-1}\) at fixed deformation temperature of 350°C, the feature of flow stress curves switched to the second case that flow stress kept steady at some certain levels along with strain, which indicated the achieved dynamic equilibrium between work hardening and softening [8]. The third case that flow stress after yield plateau declines along with strain is associated with the fact that the rate of dynamic softening is enhanced and are already exceed the rate of dynamic hardening. The enhanced dynamic softening could be resulted from thermal softening [12]-[14], dynamic recovery (DRV) [15], dynamic recrystallization (DRX) [16]-[18] and coalescence of precipitates [19], [20].

Thermal softening usually occurred when the deformation was conduct at high strain rate [12]-[14]. Shi reported the thermal softening phenomena in 7150 aluminium alloys [14]. When a 10s\(^{-1}\) strain rate was set, the deformation was finished within only 0.08 second. Such a short deformation time resulted that the heat produced from deformation could not fully released. Hence an as high as 35 °C temperature rise was induced in the tested sample, led to a thermal softening mechanism during hot deformation.

DRX in the microstructures was an important softening mechanism during hot deformation. DRX could often be triggered when deformation was performed at high temperatures and low strain rates [14]. During deformation, dislocations and subgrains were progressively developed inside the deformed grains, the slow straining process and high temperature allows the accumulated substructures progressively converted into grain boundaries via the combining of dislocations and rotation of subgrains [5], [21]. The newly formed grains are featured as free of substructures and are much softer than the original deformed ones, leading to the dynamic softening during hot deformation. [22], [23].

The softening mechanism of coalescence of precipitates involves the loss of strengthening effect exerted by the coarsen or the dissolution of fine precipitated particles [19]. During hot deformation, once coalescence of precipitates occurred, the strong pinning effect of fine precipitates on dislocation movement was severely weakened, leading to enhanced rate of dynamic softening. Besides, the coarsening and/or dissolution of precipitates further improves the rate of DRV/DRX, which resulted in a more remarkable dynamic softening [24], [25].

The analysis of flow stress curves is a very classic topic in the field of aluminium research. A series of investigation, including microstructural observation, data-based analysis, modeling and predicting, have been carried out by researchers [26]-[29]. The investigation specifically on the novel dispersoids-strengthened 6082 alloys still cannot found in the open literatures. The mechanism behind the changes in flow stresses of novel dispersoids-strengthened 6082 alloys need to be better understood.

In the present study, flow stress curves of base 6082 alloy free of Mn and a novel dispersoids-strengthened 6082 alloy were obtained by performing hot-compression tests with temperatures ranged from 400 to 550 °C and strain rates from 1 to 0.001 s\(^{-1}\). The hot deformation flow stress curves were analyzed, typical microstructural evolution during deformation was examined. The focus of this study is to understand the main mechanism of flow stress behavior in the novel dispersoids-strengthened 6082 alloy.

2. Experimental methods

Two 6082 type alloys, base alloy free of Mn and 0.75Mn alloy with a 0.75% Mn addition, were used for the investigation. The chemical compositions are shown in Table 1. The billets were cast prepared using direct chill cast with 101 mm diameter. Heat treatment of 450 °C for 6 h with a 100°C/h heating rate was performed in order to promote the precipitation of dispersoids in the aluminium matrix [4]. Cylindrical specimens with diameter 10 mm and length 15 mm were prepared for uniaxial hot-compression tests. The hot-compression were performed on a Gleeble 3800 thermo-mechanical
The specimens were heated at 2 °C/s and held to the desired temperatures for 180 s to ensure a homogeneous temperature distribution. The specimens were then compressed to a total true strain of 0.75 with the temperatures ranged from 400 to 550 °C and strain rates from 1 to 0.001 s⁻¹. The flow stress date was obtained by the testing unit and then saved automatically. A series of flow stress curves were drawn in function of true strain and true stress at different deformation temperatures and strain rates. For the observation of microstructures, an optical microscope (Nikon, Eclipse ME600), a scanning electron microscope (SEM, JEOL-6480LV) and a transmission electron microscope (TEM, JEM−2100) operating at 200 kV were applied. An electron back-scattered diffraction (EBSD) analysis was used to check the grain structures. All Euler orientation maps were applied in the EBSD image analysis. Moreover, a quantitative analysis of the dispersoid particle was performed based on the SEM images of the polished sample surfaces, some of the samples were etched with a 0.5% HF solution for 40 s to further reveal the detail of microstructures. A Sigmascope SMP10 eddy-current device was used at room temperature to measure the electrical conductivity to follow the solid solution levels.

| Table 1. Chemical compositions (wt.%) of the experimental alloys. |
|------------------|---|---|---|---|---|
| Alloys    | Mg | Si | Fe | Mn | Al |
| Base      | 0.79 | 1 | 0.18 | - | Bal. |
| 0.75Mn    | 0.84 | 1.02 | 0.23 | 0.72 | Bal. |

3. Results

3.1. Hot deformation flow stresses
Hot deformation was conduct at temperatures ranged from 400 to 550 °C and strain rates from 1 to 0.001 s⁻¹. Figure 1 shows a series of deformation true stress–true strain curves obtained under the given experimental conditions. For all the deformation conditions, the flow stress raised sharply at the beginning of the deformation until reaching peak stresses. Subsequently, 3 types of trend in flow stress curves, continuous but slow increase, remaining stable at certain levels and decrease smoothly with increasing strain, were observed, depending on the different alloys and deformation parameters.

The continuous increase of flow stresses after yield were mainly presented in the base alloy when the strain rate was 1s⁻¹ at all four deformation temperatures as well as in the 0.75Mn alloy at 1s⁻¹ strain rate under the 400 and 450 °C deformation temperatures. For the case of flow stress remained stable at a certain level after peak, the typical conditions were presented in the base alloy when the deformation strain rates of were 0.1, 0.01 and 0.001s⁻¹ at all four deformation temperatures (Figure. 1a, c, e). The decrease of flow stress after peak with increasing strain could be only distinguished in the 0.75Mn alloy, the corresponding conditions covered nearly all the deformation regimes except that of the 1s⁻¹ strain rate at 400 and 450 °C (Figure. 1b, d, f).

Besides, some evident general tendencies were observed, i.e., the level of flow stress decreased with the increase in deformation temperatures and increased with the increase in strain rates for both base and 0.75Mn alloys. Additionally, the overall flow stress levels in the 0.75Mn alloy were higher than those of base alloy under same deformation parameters.

To further demonstrate the flow stress of two experimental alloys, peak flow stresses in both alloy and flow stress at 0.75 strain in 0.75Mn alloy were chosen as the typical values to clarify the features of flow stress curves. As shown in Figure. 2, in general, overall higher peak flow stresses in 0.75Mn alloy (Figure. 2b) than those of base alloy (Figure. 2a) were distinguished. The peak flow stress increased with increasing strain rate and decreasing temperature. Flow stress at 0.75 strain in 0.75Mn alloy showed the similar tendencies as those of peak flow stress, while an extra feature was that the decrease of stress with decreasing strain rate was more remarkable than those of peak stress (Figure. 2b).
In order to further characterize the differences in the peak flow stress and flow stress at 0.75 strain in 0.75Mn alloy, a decline ratio of flow stress, $R_d$, is introduced and defined as Eq. (1) [7]:

$$ R_d = \frac{\sigma_p - \sigma_{0.75}}{\sigma_p} \times 100\% $$

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**Figure 1.** True stress–true strain curves of hot compression deformation, (a), (c), (e) and (g) base alloy; (b), (d), (f) and (h) 0.75Mn alloy.
Figure 2. Peak flow stresses in base alloy (a), peak flow stress and stress at 0.75 strain of 0.75Mn alloy (b) at different hot deformation temperature as a function of the strain rate.

Where $\sigma_p$ is the value of peak stress and $\sigma_{0.75}$ is the value of flow stress at the 0.75 strain of the deformation. Figure 3 shows the results of $R_d$ of 0.75Mn alloy in function of strain rate at different deformation temperatures. In general, $R_d$ decreased with increasing strain rate and decreasing temperature. Specifically, when the deformation strain rate was 1s$^{-1}$, -8.3% at and -4.3 of $R_d$ value were derived from 400°C and 450°C deformation data respectively. These negative values respected the increase of flow stress along with strain after yield plateau (Figure. 1). As the deformation temperature increased by 500 and 550°C at the 1s$^{-1}$ strain rate, $R_d$ became greater than 1, corresponded with the drop tendency of flow stress curve after yield plateau at these temperatures. Regarding the strain rates of 0.1, 0.01, 0.001s$^{-1}$, $R_d$ values for all four deformation temperatures became positive value, which was associated with the increased flow stress curves in all four deformation temperatures.

Figure 3. Decline ratio of flow stress of 0.75Mn alloy in function of strain rate at different deformation temperatures.

3.2. Microstructures evolution during deformation

The experimental 0.75Mn alloy was designed as a dispersoid-strengthening alloy. Addition of 0.75% Mn and the heat treatment ensured the fully precipitation of dispersoids to promote the high temperature mechanical properties [4], base alloy free of Mn was prepared to make a comparison. Figure. 4 shows the microstructures of the two alloys after deformation of 450 °C/ 0.1 s$^{-1}$. Mainly elongated grains were observed in the samples after hot deformation. Fe-rich intermetallics (white in the inset SEM images) and primary Mg$^2$Si particles (black) were distributed along the dendrite boundaries (Figure. 4a, b). On the surface of etched samples, the dispersoids can be clearly revealed in 0.75Mn alloy (Figure. 2b, c), these dispersoids were produced through the Mn addition and the 450°C/6h heat treatment. Our previous study had confirmed these dispersoids as $\alpha$-Al (Fe, Mn) Si phase [4]. SEM image (Figure. 4c) shows that the dispersoids locate at the region inside the dendrite cells. Meanwhile, dispersoids free zone formed near the dendrite boundary regions where
intermetallics was observed. The morphology of dispersoids was closely observed in TEM image (Figure, 4d), they appeared mostly in round or elliptical shapes in the two-dimensional image. Besides, dislocations interacted with the dispersoids were observed, indicating a retarding effect of dispersoids on dislocation motion during deformation.

![Image](image_url)

**Figure 4.** Microstructures of base alloy and 0.75Mn alloys after 450 °C/0.1 s⁻¹ deformation: SEM images of etched surface (with non-etched image inset) of (a) base alloy and (b) 0.75Mn alloy, enlarged SEM image of etched surface of 0.75Mn alloy (c), (d) bright field TEM image of 0.75Mn alloy.

The dispersoids number density before and after deformation were quantified base on the SEM images. The results suggested that 0.75Mn alloy had a dispersoids number density of 11.5μm⁻² before deformation. After deformation was performed, the change of dispersoids number density was fairly limited at most of deformation conditions, while only after 550°C/0.001s⁻¹ deformation, an obvious decrease in the population of dispersoids was observed. Figure. 5 shows the typical etched surface of 0.75Mn alloy after 450°C/0.1 s⁻¹ and 550°C/0.001s⁻¹ deformation. In the Figure. 5a, the dispersoids number density is 11.5μm⁻², which is close to that of before deformed samples, while in Figure. 5b a much small population of dispersoids (dispersoids number density dropped by 5.6 μm⁻²) was observed, suggesting the dissolution of dispersoids took place during the high temperature-low strain rate deformation.

Evolution of primary Mg₂Si particles were followed. At the low deformation temperatures range of 400-500°C, no occurrence of Mg₂Si dissolution was observed for all four strain rates. At 550°C of deformation temperature, strain rates no less than 0.01s⁻¹ still was still not able to make full dissolution of Mg₂Si (Figure. 6a, c) in both base and 0.75Mn alloys. Only the lowest strain rate of 0.001s⁻¹ with highest deformation temperature of 550°C could lead the full dissolution of Mg₂Si (Figure. 6b, d).
Grain structures were investigated using EBSD technology. Figure 7 shows All Euler orientation maps of two alloys under the three typical deformation conditions. In the orientation maps, different colors represent the orientation differences of grains and subgrains. Elongated grains perpendicular to the compression direction were observed in all deformed samples. At the 450 °C/0.1 s⁻¹ deformation condition, a large amount of white and green lines was observed in the elongated grains, implying the high densities of dislocation cells and subgrains (Figure 7a, d) in both base and 0.75Mn alloys. These EBSD map were typically characterized as dynamic recovery (DRV) [5]. When the deformation temperature raised to 550°C at 0.1s⁻¹ strain rate, the density of dislocation cells and subgrains were obvious lowered (Figure 7b, e), the reduction of substructures in base alloy was much evident than that in 0.75Mn alloy. Besides, some grains nearly free of internal substructures were observed in base alloy (arrows in Figure 7b), indicating occurrence of dynamic recrystallization (DRX) [5], [21], [22]. However, DRX was not observed in 0.75Mn alloy, DRV was still the predominant mechanism after deformation. When the strain rate decreased by 0.001s⁻¹ at 550°C deformation temperature, a nearly full DRX took place in base alloy, reflected by the almost disappearance of substructures (Figure 7c). In 0.75Mn alloy, some small grains free of internal substructures were observed near the original grain boundaries (arrows in Figure 7f), suggesting the occurrence of partial DRX during hot deformation.

Figure 5. Dispersoids in 0.75Mn alloy after (a) 450°C/0.1 s⁻¹ and (b) 550°C/0.001s⁻¹ deformation.

Figure 6. SEM image after deformed at 550°C with strain rate of (a) 0.01 s⁻¹ in base alloy, (b) 0.01 s⁻¹ in 0.75Mn alloy, (c) 0.001 s⁻¹ in base alloy and (d) 0.001 s⁻¹ in 0.75Mn alloy.
Figure 7. All Euler orientation maps of two alloys under different deformation conditions. The boundary misorientation angles are marked by white lines 2-5°, green lines 5-15° and black lines > 15°. A step size of 0.5 μm was set during the orientation map scanning.

4. Discussion
Hot deformation flow stress and the microstructures evolution were investigated based on 6082 base alloy and a novel dispersoids-strengthened 0.75Mn 6082 alloy. The hot deformation flow stress curves exhibited various features, which was closely related to deformation parameters and the evolution of microstructures reflecting the dynamic balance between hardening and softening.

4.1 Increased flow stress after yield plateau
The continuous increase of flow stresses after yield plateau were mainly presented in the base alloy when the strain rate was 1s⁻¹ at all four deformation temperatures as well as in the 0.75Mn alloy at 1s⁻¹ strain rate under the 400 and 450 °C deformation temperatures [Figure. 1 and 3]. During deformation process, dislocations were introduced progressively. With increasing strain, the accumulation and pile-up of dislocation became more severe, resulting in the high rate of dynamic work hardening [30]. Meanwhile, the occurrence of dynamic softening via the combining and rearrangement of dislocations and subgrains, which was associated with DRV, could partially release the accumulated energy and lead to a weakening of work hardening [31]. In case of high strain rate of 1s⁻¹, the while duration of deformation was only 0.8 second, in such a short time, DRV was not able to take place in highly active degree, corresponding dynamic softening rate stayed at a lower level than that of work hardening and was responsible for the continuous increasing flow stress along with strain.

4.2 Steady flow stress level after yield plateau
Steady flow stress level after yield plateau was observed main in the base alloy when the deformation strain rates of were 0.1, 0.01 and 0.001s⁻¹ at all four deformation temperatures [Figure. 1]. The steady flow stress implied that an equilibrium balance between dynamic hardening and dynamic softening was reached and was maintained until the end of deformation. When the deformation was performed under lower strain rate, the longer duration allows the high degree of DRV and DRX take place (Figure. 7b, c), resulting in the enhanced dynamic softening. when an equilibrium balance between dynamic hardening and dynamic softening was reached, the flow stress curves exhibited the steady feature. The similar steady flow stress was also observed in the dilute aluminium alloy [32] and some Al-Mg-Si alloys [27], [33], [34].
4.3. Decreased flow stress after yield

The decrease of flow stress after yield plateau with increasing strain was presented only in the 0.75Mn alloy. The corresponding conditions covered all the deformation regimes except that of the 1s\(^{-1}\) strain rate at 450 and 500°C.

Decline of flow stress after yield plateau was a sign that the dynamic softening rate had exceed that of dynamic hardening. It was reported in literatures that the dynamic softening could be associated with coalescence of precipitates [19], [20], dynamic recrystallization (DRX) [12]-[14] and thermal softening [12]-[14].

In 0.75Mn alloy, the overall higher flow stress level than those in base alloy could be the result of the strong strengthening effect of α-Al (Fe, Mn)Si dispersoids (Figure. 1, 2) [4]. Dispersoids number density was an important parameter influencing flow stress levels and contributing the change in flow stress curves. It is reported that [35] strengthening of α-Al(Fe, Mn)Si dispersoid is the predominant strengthening mechanism for a Al–Mn–Mg–Si alloys. In this study, the decline of flow stress after yield can be interpreted as being due to the dissolution of dispersoids. The dispersoids number density before deformation in 0.75Mn alloy was 11.5μm\(^{-2}\) (Figure. 4a), whereas after deformation at 550°C with 0.001s\(^{-1}\) strain rate, the dispersoids number density dropped by 5.6μm\(^{-2}\) (Figure. 4b), indicating that a loss of 49% population of dispersoids during the deformation, which may be responsible for the severe drop in flow stress from 20.7MPa at peak by 9.6MPa at 0.75 strain. However, for all of the rest condition of 0.1s\(^{-1}\), 0.01s\(^{-1}\) and 0.001s\(^{-1}\) strain rate at all four deformation temperatures, where no sever dissolution of dispersoids took place, still it could be observed the decline of flow stress after yield (Figure. 1 and Figure. 3), which implying a fact that the dissolution of dispersoids was not the direct reason responsible for the declining flow stress.

In 0.75Mn alloy, DRX was observed in some deformed samples, for example, 550/0.001s\(^{-1}\) in 0.75Mn alloy. It is reported that, in Al–Mg–Si–Cu aluminium alloy, when the DRX occurred, an obvious decline or fluctuation in flow stress was observed due to the releasing of stored strain energy via formation of new grains [26, 30]. Wherever, in 0.75Mn alloy no evident flow stress fluctuation along with strain could be observed, all the flow stress curves exhibited quite smooth decline rates. Besides, under the deformation conditions where no DRX occurred (for example, 550°C/0.1s\(^{-1}\) condition), still the decline of flow stress after peak plateau was observed, indicating that DRX may be not the key factor determining the decline of flow stress.

Thermal softening was reported to be a softening mechanism under the conditions when the deformation was conduct at high strain rate [12]-[14]. In 7150 aluminium alloys [14], When train rate was set as 10s\(^{-1}\)s, a 35 °C temperature rise was induced in tested samples due to the strain heat could not fully released within as short as only 0.08 second of deformation duration. This raised temperature resulted in a thermal softening mechanism during hot deformation. However, in the present study, the highest strain rate is 1s\(^{-1}\), no evident temperature raise in deforming specimen was tested, which ruled out the possibility of thermal softening.

The above mechanisms of dispersoids dissolution, DRX mechanism and thermal softening seem to be difficult to clearly explain the decline of flow stress after yield plateau in 0.75Mn alloy. A solution diffusion rate-controlling mechanism that involves the changes in the solid solution strengthening may give an explanation for the decline of flow stress phenomenon in 0.75Mn alloy.

On view of solutions strengthening in the Al matrix, changes in solution levels can significantly influence the hot deformation flow stress of aluminium alloys [36], [37]. Shakiba studied the hot workability of dilute Al-Fe-Si alloys and found that increasing the iron solution level from 0.1 to 0.7% produced 11 to 32% increase in the hot deformation flow stress in a dilute Al-Fe-0.1Si alloy due to the solid solution strengthening effect of Si and Fe elements [11]. The increased flow stress level resulted from solid solution level was also reported in the 1xxx alloy during homogenization [32]. A rise in the homogenization temperature from 550 to 630 °C resulted in the increased flow stress by 15–45% in a 1xxx alloy due to the enhanced release of solution element to the matrix at high homogenization temperatures.

In the present study, Mg, Si, Fe and Mn was involved in view of solid solutions. Technically, it is not an easy work to directly test the solid solution levels of element in the aluminium matrix. In this case,
electrical conductivity is usually used as an indicator to follow the change in solid solution level [32]. In this study, the electrical conductivity of base and 0.75Mn alloy after heat treatment (before deformation) were 48.9%IACS and 47.4%IACS respectively. This lower electrical conductivity in 0.75Mn alloy implied that, despite a 450°C/6h heat treatment was conduct to promote the precipitation of Mn-containing dispersoids, there still existed a certain amount of Mn remained in Al matrix. This extra Mn solute present in 0.75Mn alloy could pose a significant solid solution strengthening effect during hot deformation.

The impact of different types of solute elements on hot deformation flow stress behavior was studied by Shakiba et. al. It is claimed that the solute diffusion rate in aluminium alloys was the key factor determining the flow stress levels of hot deformation. When different types of solution elements were individually added in the aluminium alloys, the hot deformation flow stress level is inversely proportional to the diffusion coefficient of the added solute elements, which is referred as solution diffusion rate-controlling mechanism [38].

Figure 8 shows the diffusion coefficient of Mg, Si, Fe and Mn in aluminium as well as that of Al self-diffusion [39]. Mg and Si have the slightly higher diffusion coefficients in Al than the Al-self diffusion, while Fe have moderately lower diffusion coefficients than those of Al-self diffusion. Mn exhibited the lowest diffusion coefficients at the given temperature range. According to the rate-controlling mechanism, Mn, which have lowest diffusion coefficients, should has the much stronger impact on the flow stress than those of Fe, Si and Mg.

For the case of Mn solute in 0.75Mn alloy, Mn solute atoms had a strong solid solution effect because of its low diffusion rate in aluminium according to the diffusion rate-controlling mechanism. At the early stage of deformation, as soon as the elastic deformation was started, the dislocations begin to produced. In this case, Mn solute atoms act as the barrier for the dislocation motion, which resulted in a high deformation resistance and lead to a hardening effect [38]. As the strain goes on, until the work hardening and dynamic softening meet at a balance rate, the yield point was reached.

After peaks, flow stress begins to decline after peak in 0.75Mn alloy, which suggested that the solution strengthening may be influenced by other factors. A “pipe diffusion” mechanism may explain the change in the solution strengthening along with strain. “Pipe diffusion” refers a phenomenon that when carried by dislocations, the solute atoms diffusion switched from the static diffusion to dynamic diffusion mechanism, so that the diffusion rate increased remarkably. Dislocations served as fast diffusion paths for solute atoms diffusion in the matrix, thereby permitting rapid dynamic diffusion of solution atoms [40], [41]. In 0.75Mn alloy, the dislocations were produced dynamically in a considerable high rate along with the strain. When the dislocations encountered Mn atoms, a “pipe diffusion” mechanism was turned on. It is reported that after the “pipe diffusion” take place, the diffusion rate became as high as several order as that at static diffuse condition [41]. In case of significantly increased Mn atom diffusion rate, the deformation resistance was weakened according to the diffusion rate-controlling mechanism. When the “pipe diffusion” take place, Mn solute atoms
somehow could not act as the barrier of dislocation motion any longer. Furthermore, when the unpinned dislocations glided passing the static Mn atoms in the matrix, static Mn atoms were “wiped out” by the dislocations and start the “pipe diffusion”, correspondingly resulting in the further severe loss of strength of the alloy. Therefore, the diffusion rate-controlling mechanism and the “pipe diffusion” seems to give a reasonable explanation for the decline of the flow stress in 0.75Mn alloy. In base alloy free of Mn, the steady flow stress after yield could also be explained by the diffusion rate-controlling mechanism and “pipe diffusion”. According to the diffusion rate-controlling mechanism, the diffusion coefficients of Mg and Si are much higher than those of Mn and was very closed to those of Al-self diffusion. This suggested that the impacts of Mg and Si in solution were much weaker than those of Mn and even approximately equaled to the Al matrix. After yield plateau, when the “pipe diffusion” was highly active, the impact of strength driven from Mg and Si solution strengthening was further weakened. Hence, no evident loss in strengthen could be produced after the turn on of “pipe diffusion” in Mg and Si, explaining the fairly steady flow stress along with strain until the end of deformation. Similar steady flow stress feature along with strain was also seen in the Al-Mg-Si alloys in the works of [27], [32], [33], in which the element contents of studied alloys were similar as those in base alloy of this study.

Fe is another element added in both alloys which has a moderately low diffusion coefficient. Theoretically, Fe has lower diffusion coefficients and should have an evident effect on the flow stress. However, the solute ability of Fe in Al matrix was quite limit (about 0.003% at 450°C) [32] after the heat treatment at 450°C for 6h. In the case of such low level of Fe element in the matrix, it is difficult for Fe to make a strong impact on the flow stress.

To sum up, the decline of flow stress after peak in 0.75Mn alloy was due to the significantly increased diffusion rate of Mn solution via “pipe diffusion”, which lead to the loss of Mn solid solution strengthening along with strain. The effect of Mg and Si played a limit role on the declined flow stress owing to their high diffusion coefficients. Fe could not provide a strong influence on the flow stress because of its low solubility.

Another phenomenon of the dissolution of primary Mg$_2$Si was observed in both base and 0.75Mn alloy at only 550°C/0.001s$^{-1}$ deformation condition. It is reported that [42], [43] fully dissolution of primary Mg$_2$Si in Al-Mg-Si alloys made the Mg and Si atoms were released to the Al Matrix, led to a significant increased flow stress level due to an improved solid solution strengthening. However, in present work when dissolution of primary Mg$_2$Si was observed under 550°C/0.001s$^{-1}$ deformation condition no evident sing of increase nor even a compensation of the decrease in flow stress after yield along with strain was observed. According to the diffusion rate-controlling mechanism and “pipe diffusion”, as soon as the Mg and Si released into the Al matrix, their diffusion was switched to “pipe diffusion” and loss their pinning ability on the dislocations motion immediately. Therefore, the impact of dissolution of primary Mg$_2$Si on flow stress was not significant.

5. Conclusions

The hot deformation tests were carried out on the base and novel dispersoids-strengthened 0.75Mn 6082 alloys in a temperature range of 400–550 °C and strain rate range of 0.001s$^{-1}$-1s$^{-1}$, a series of flow stress curves were obtained. Prior to the hot deformation, a heat treatment of 450 °C for 6 h was conduct, a large numbers of α-Al(Fe, Mn)Si dispersoids were precipitated in 0.75Mn alloy. Conclusions were drawn as followed:

(1) Deformation flow stress levels increased with increasing strain rate and decreasing temperature, 0.75Mn alloy had overall higher flow stress level than those of base alloy.

(2) For base alloy, when deformed at 1s$^{-1}$ strain rate, flow stress curves showed continuous increase after the yield plateau, while at 0.1, 0.01 and 0.001 strain rates flow stress remained at certain levels at all four deformation temperatures of 400, 450, 500 and 550°C.

(3) Regarding the 0.75Mn alloy, the increase of flow stress after yield plateau was observed only under the 1s$^{-1}$ strain rate at 400 and 450°C of deformation temperatures. For the rest of deformation conditions, including 1s$^{-1}$ strain rate at 500 and 550°C and 0.1, 0.01, 0.001s$^{-1}$ strain rates at all four deformation temperatures, the flow stress after yield plateau exhibited the decline tendency.

(4) Under 550°C/0.001s$^{-1}$ deformation condition, full dissolution of primary Mg$_2$Si occurred in
both alloy, dissolution of dispersoids took place in 0.75Mn alloy. Partial DRX was observed in base alloy at 550°C/0.1s\(^{-1}\) deformation condition. At 550°C/0.001s\(^{-1}\) deformation condition, nearly full DRX in base alloy and partial DRX in 0.75Mn alloy was recognized.

(5) Neither the dissolution of dispersoids nor the DRX was responsible for the decrease of flow stress in 0.75Mn alloy. The decreased flow stress was explained using a solution diffusion rate-controlling mechanism. When diffusion of Mn solution was active via “pipe diffusion”, the diffusion rate of Mn increased significantly and lead to the decrease in flow stress.

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