

Article

Effect of Precipitation Behavior on Mechanical Properties of 6082 Aluminum Alloy

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Abstract

The mechanical properties of 6082 aluminum alloy under different aging conditions were investigated in this study. Precipitation behavior leading to an increase in strength was confirmed by the observation of the microstructure using TEM. The β'' precipitate, which is coherent with the Al matrix, contributes to the high tensile strength and hardness of the alloy. The transformation of the β'' phase into the β' phase observed under TEM was proven to be responsible for the decrease in strength and hardness. The hardness curves indirectly reveal that a higher aging temperature could accelerate the transformation of β'' into β' . The coherent strain field introduced by the β'' precipitate was clearly observed and is discussed. In addition, an ideal artificial aging window is proposed, with a combined performance of a TS of 360 MPa, hardness of 105 HB and elongation over 10%.

Keywords: aluminum alloy 6082; artificial aging; β'' precipitate; precipitation hardening; TEM



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1. Introduction

With the increasing demand for weight reduction and fuel saving in the transportation field, lightweight bodies have been increasingly used in automobiles. Aluminum alloy is selected as a reasonable material due to its low density and ease of recycling [1]. Compared to other lightweight materials, aluminum alloy is considered as an excellent substitution for steel to satisfy the requirements of being lightweight, as its density is just about one-third of that of steel. Due to the excellent characteristics of high strength, good plasticity, corrosion resistance, and good stamping performance, Al-Mg-Si alloy has received more and more attention in the automotive industry. Beyond the inherent lightweight advantages of the material itself, optimizing structural design can further advance lightweighting goals. Extruded aluminum alloy extruded profiles, such as tubing, are gaining increasing attention in the automotive industry. The final properties of the 6082 Al alloy extrusion profile are determined by the heat treatment process, including solid solution and aging treatments. To satisfy the different design requirements, the influence of the heat treatment on Al-Mg-Si alloy needs more attention.

Precipitate hardening contributes to alloy strengthening [2]. The main precipitates of 6082 aluminum alloy are Mg_2Si , the quantities and distribution of which have a significant effect on the alloy properties. The contents of Mg and Si are usually in the range of 0.5–1 wt%, and with the beneficial effect obtained from an excess of Si, the Mg/Si ratio is generally lower than 2. Mn, as an important additional element, is about 0.7 wt% and can effectively reduce damage caused the impurity Fe, the content of which is about 0.2% [3].

The precipitation sequence of Al-Mg-Si alloys is generally accepted as follows [4–6]: α -supersaturated solid solution $\rightarrow$ atomic clusters $\rightarrow$ vacancy-rich G.P.-I zones (Guinier–Pereton zone) $\rightarrow \beta''$ (G.P.-II zones) coherent precipitates $\rightarrow \beta'$ semi-coherent precipitates $\rightarrow \beta$ (Mg_2Si) incoherent precipitates.

The microstructural morphology of the hardening precipitates can strongly influence the magnitude of the strengthening by the strain energies and interface between precipitates and the Al matrix [7]. Technically, Al-Mg-Si alloy can obtain a peak hardness with the collective effect of full fine G.P. I zones and β'' coherent precipitates [8]. G.P. zones, derived from an Al lattice by replacing the Al atoms with Mg and Si atoms, are generally considered to be the precursors for the major strengthening phase β'' , and extra GP zones would serve as nuclei for β'' precipitates.

In general, the improvement in strength is associated with the stage of the precipitate. Needle-like β'' precipitates, of which the composition is Mg_5Si_6 , are considered as the direct cause for peak-aged states [9–11]. These hardening phases are fully coherent along the needle axis and aligned as $\langle 100 \rangle_{\text{Al}}$.

The precipitate behavior of Al-Mg-Si alloy remarkably depends on a combination of the aging temperature and aging time. The required properties of 6082 aluminum alloy can be obtained with a reasonable heat treatment procedure. Yang et al. found that for 6082 aluminum alloy, the hardness first increases and then decreases when the aging time is 1 h [12]. In contrast, Bartawi observed that the peak hardness of Al-Mg-Si alloys appears after 5 h of aging [13]. On the other hand, Hu investigated the effects of strain rate and aging time on 6082 aluminum alloy; they found that dislocations introduced by deformation accelerate the precipitation kinetics of precipitates, which shortens the peak aging time and promotes the formation of composite phases [14]. Jiang et al. studied the influence of conventional T6 heat treatment on 6082 aluminum alloy flange components and proposed an optimal T6 heat treatment process [15].

It can be observed that the research on 6082 aluminum alloy has limitations, in that it focuses solely on aging temperature, aging time, or specific components. This work aims to investigate the relationship between the precipitates and the properties of automobile component made of a 6082 extrusion profile, explain the essential reasons for the changes in mechanical properties corresponding to the precipitate behavior, and put forward reasonable aging processes for industry practice.

2. Materials and Methods

This work was carried out on a commercial 6082 aluminum alloy extrusion profile. Table 1 lists the chemical composition of the alloys.

Table 1. The chemical composition of the investigated alloys (wt%).

Alloy	Si	Mg	Fe	Cu	Mn	Cr	Al
6082	0.85	0.80	0.21	0.05	0.50	0.06	Bal.

2.1. Heat Treatment Scheme

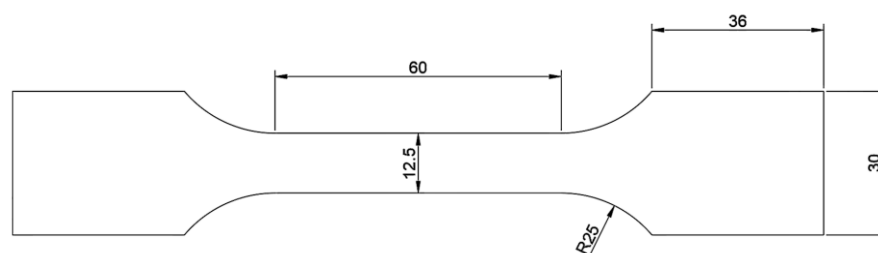
According to the literature review, the alloy underwent a 2 h solution treatment at 530 °C, followed by immediate water quenching to room temperature. Then, the alloy was heated to an aging temperature ranging from 90 °C to 190 °C and then held for 1–10 h, as shown in Table 2. The influence of the aging treatment parameters on the precipitation mechanisms and kinetics of the 6082 alloys was investigated.

Table 2. Heat treatment parameters.

Solution Treatment	Aging Temperature/°C	Aging Time/h
530 °C + 2 h	90/140/170/190	1/2/4/6/8/10

2.2. Determination of Mechanical Properties

To determine the overall performance response of the specimens, the Brinell hardness was selected as the test criterion. An HBE 3000A-type hardness tester manufactured by Shanghai ShuangXu Electronics (Shanghai, China) was used to measure the hardness of the samples with a load of 62.5 kg and a dwell time of 30 s. According to the standard GB/T228.1-2010 [16], a tensile test was conducted to examine the mechanical properties of the specimens using the CMT5305 tester manufactured by MTS (Eden Prairie, MN, USA) at a constant displacement rate of 3 mm/min. Figure 1 shows the size of the tensile sample.

**Figure 1.** Diagram and dimensions of tensile test specimens.

2.3. Metallographic Investigations

Fracture morphology examination of the tensile test specimens was carried out using a SUPRA 40 scanning electron microscope (SEM) manufactured by ZEISS (Oberkochen, Germany). The specimens were first ultrasonically cleaned and dried before the observation.

The microstructures of the specimens were observed using a JEM 2010F transmission electron microscope (TEM) manufactured by JEOL (Tokyo, Japan), operating at 200 kV. Before the TEM observation, the sample was mechanically thinned to 60 µm and then electropolished using a solution of 10 vol% perchloric acid in alcohol at −30 °C with a current of 30 mA using a twin-jet machine.

3. Results

3.1. Mechanical Properties

The results of the tensile test and Brinell hardness test are given in Figures 2–4.

3.1.1. The Ultimate Tensile Strength (TS) and Total Elongation (TE)

Figure 2 is based on the mechanical properties of the 24 samples, each one of which was treated under 1 of the 24 different kinds of heat treatments, combining four aging temperatures and six aging hours. As Figure 2 shows, the aging temperature strongly affected the tensile strength. With the same aging time, the tensile strength increased with the increase in aging temperature, and a peak value occurred at around 170 °C. After the peak value temperature (170 °C), the TS did not increase with the temperature any further. In addition, the tensile strength decreased slightly with the increase in aging temperature, with the ultimate TS of the 190 °C sample being lower than that of the 170 °C sample. According to Figure 2, the strength of the specimen reached over 300 MPa when the Al-Mg-Si alloy was heated to a temperature ranging from 160 °C to 190 °C.

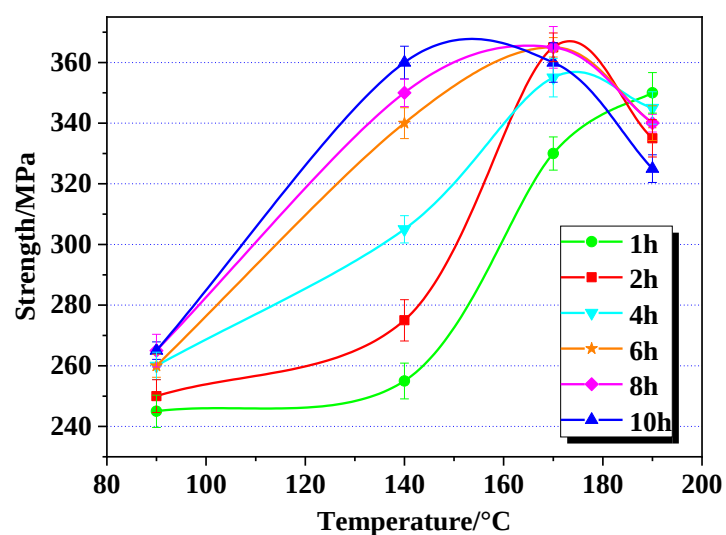
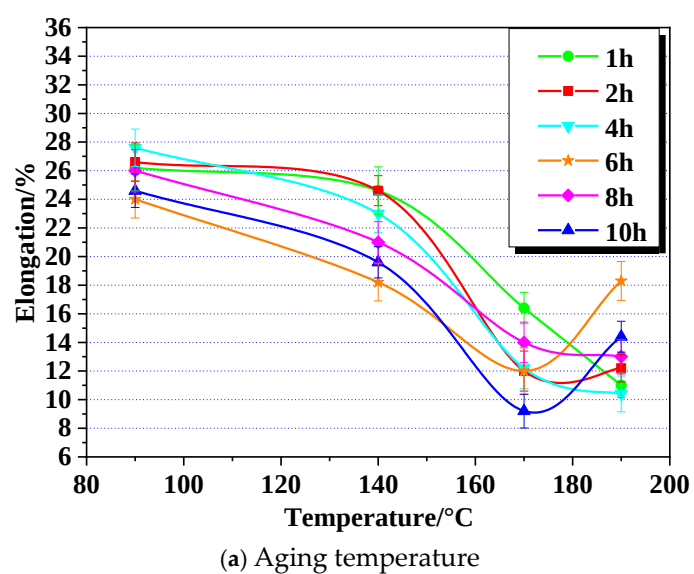
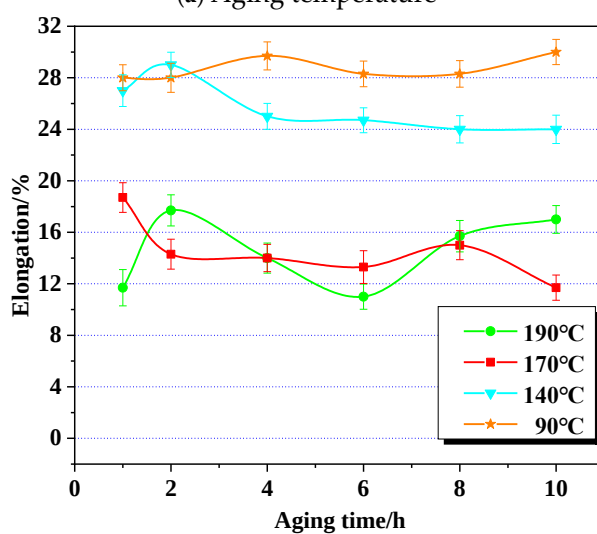


Figure 2. Effects of aging procedures on TS of 6082 alloy.



(a) Aging temperature



(b) Aging time

Figure 3. Effects of aging procedures on total elongation of 6082 alloy.

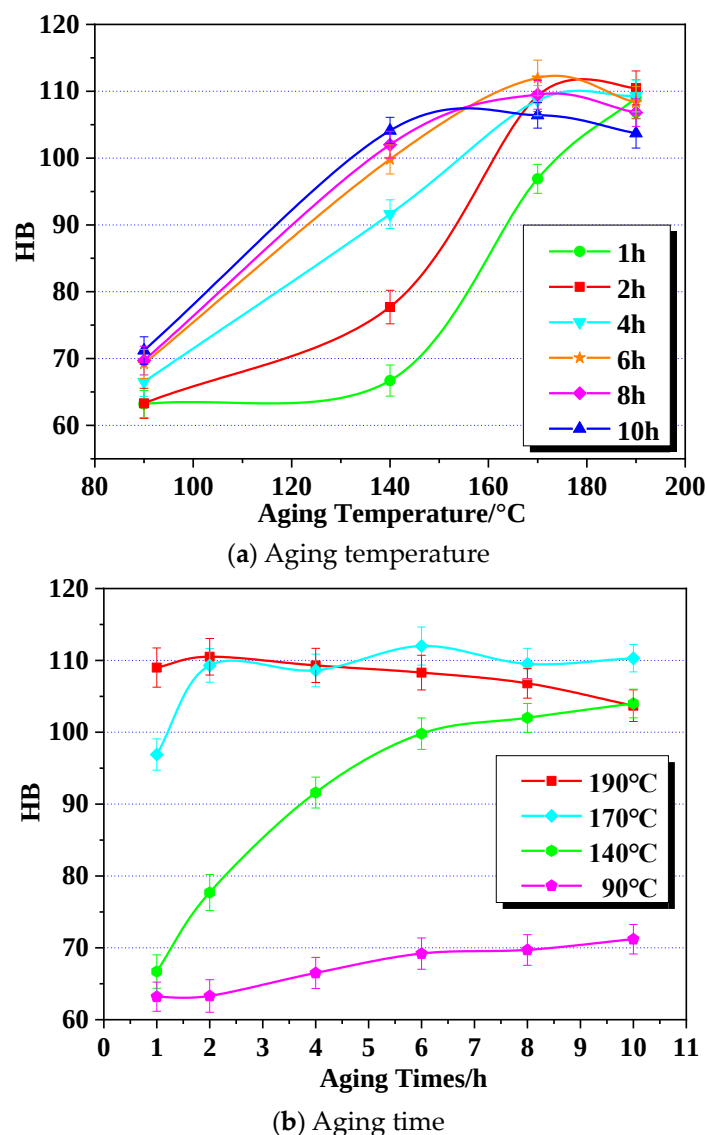


Figure 4. Effects of aging procedures on hardness of 6082 alloy.

The aging time is also a key factor that is responsible for the changes in strength. As Figure 2 shows, the TS of the samples aged at 140 °C increased very sharply with time from 1 h to 10 h. This change directly relates to the precipitation behavior and indicates that the quantity of hardening precipitates becomes larger with the increase in time. Meanwhile, the ultimate tensile strength of the 170 °C sample stayed at a fairly high strength level over a wide range of aging time.

According to the gradient of each hardening line, it is obvious to see that for the same aging temperature (140 °C), the longer the aging time, the higher the strength. This shows that a high energy density with a higher aging temperature could effectively shorten the aging time.

Figure 3a shows the effect of the aging temperature on the total elongation (TE) of the 6082 alloy. There was a common phenomenon that the TE of the samples decreased with the increase in aging temperature and reached a low point at 170 °C. However, the TE of the samples increased with a higher temperature (190 °C). As a result, the aging temperature strongly influenced the elongation results, which is deeply related to the precipitation behavior. Hard precipitates could hamper the deformation of the Al matrix, leading to a relatively low elongation. The six TE curves indirectly reveal that a temperature range from 160 °C~180 °C may provide the optimal activation energy for precipitation.

As shown in Figure 3b, the elongation curves maintained a relatively steady level within the experimental time, which means that the processing time could slightly affect the plasticity. The 90 °C/140 °C samples had better performance in terms of plasticity than the 170 °C/190 °C samples. Due to the precipitation behavior, the elongation being lower at higher temperature agrees with the plastic rule.

3.1.2. Hardening Behavior

Figure 4a presents the effect of the aging procedures on the HB hardness of the experimental samples. The colored lines represent the HB curves of each procedure. The six different colored lines correspond to the six different aging time procedures (1 h/2 h/4 h/6 h/8 h/10 h), respectively, exhibiting the aging temperature–HB relationship. From 90 °C to 140 °C, obvious hardening behavior was revealed, showing that the 6082 alloy aged at low temperatures needed more time to reach a high level of hardness. A longer aging time introduced more hardening precipitates. However, the relatively high temperature range (about 170 °C) shows that using a longer heat treatment time is not helpful for the increase in HB. The hardness of the 10 h/170 °C sample was 106 HB, while that of the 6 h/170 °C sample reached 113 HB. Furthermore, after the peak value point (170 °C), a higher aging temperature (190 °C) slightly decreased the hardness of the alloy. The decrease in hardness implies the transformation of a hardening phase.

The relationship between the aging time and the hardness is shown in Figure 4b. Based on the hardness results, the effect of the aging time on the specimens aged at different temperatures is different. The 190 °C specimen reached the peak hardness with very little time, and the HB of the sample decreased slightly with the increase in aging time. Meanwhile, the 170 °C specimen maintained a high and stable HB level (about 110 HB) within the experiment time. However, the HB value of the 140 °C sample increased rapidly with the increase in aging time, which reveals that 140 °C was not able to supply sufficient energy for precipitation. Based on the hardness curve, the 140 °C sample seemed to have a further increase with a longer aging time. A temperature of 90 °C is obviously not the right processing temperature for the 6082 alloy to result in precipitation hardening. During the experimental time, the hardness of the 90 °C specimen increased slowly and maintained an HB of no more than 90.

3.2. The Fracture Morphology

By SEM observation, the fracture surface of the fractured tensile test samples under different aging conditions was investigated. To understand the influence of the peak aging value on the fracture behavior, the 90 °C/6 h sample and the 170 °C/6 h sample were selected for comparison. As the hardness of the β' and β'' precipitates is higher than that of the matrix, during uncoordinated deformation, the voids nucleated around these precipitates, and the breaking of intermetallic particles into several small pieces can induce voids. The voids introduced cracks, and the cracks propagated with the tensile stress until the sample fractured.

As is shown in Figure 5, the fracture morphologies of the two selected samples can be clearly observed and confirmed as a ductile fracture surface. The fracture morphology of the aging-treated samples mainly consists of ductile dimples with a uniform distribution and tear ridges. The dimple diameter of the 90 °C sample is about 8~10 μm , while that of the 170 °C aging sample is about 4~7 μm . The dimples of 90 °C are much deeper than those of the 170 °C sample. The shape and size of the dimples directly reflect the material's toughness. Deeper dimples means that the material is better in terms of plasticity. As the SEM observation shows, the dimples are much deeper for the sample aged at 90 °C compared to those of the sample aged at 170 °C, which matches the total elongation results well.

The enlarged fracture morphologies of the ductile dimple are shown in Figure 5c,d. It can be clearly seen that the dimple diameter of the 90 °C sample is much larger than that of the 170 °C sample, which maintains the same plasticity performance of the Al matrix. Due to the hardening precipitates hampering the deformation of the matrix and introducing voids as crack sources, the 170 °C sample fracture consists of many small-sized dimples, which are presented in the enlarged SEM results. The platform that occurs at the fracture morphology of the 170 °C sample is actually composed of smaller and denser dimples. According to the total elongation results, a conclusion can be drawn that the Al matrix mainly contributes to the plasticity of the sample, and the hardening precipitates play a role in blocking the deformation of the Al matrix.

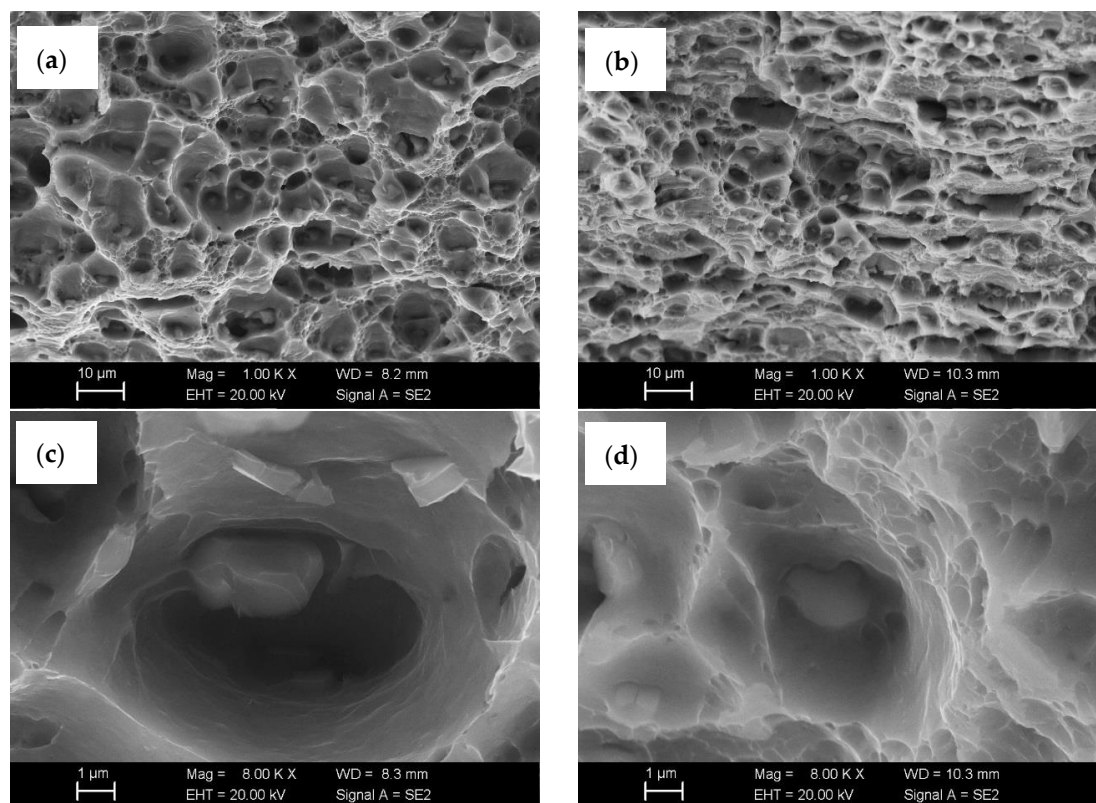


Figure 5. Fracture morphologies of specimens aged for 6 h at 90 °C (a) and 170 °C (b); enlarged figures of 90 °C (c) and 170 °C (d).

3.3. Precipitates' Morphology Observed Under TEM

The 170 °C/190 °C samples were selected for comparison to find out the effect of the heat treatment procedure on the precipitate behavior. The TEM images of the samples aged at 170 °C/190 °C for 6 h are shown in Figure 6. The β'' particles are small needles oriented parallel to the $\langle 100 \rangle$ direction in the Al matrix. Thus, TEM analysis of the precipitates was performed with the specimens in the $\langle 100 \rangle_{\text{Al}}$ zone axis orientation, which is the best orientation for visualizing the β'' phase. The β'' precipitate is considered to be responsible for the maximum age-hardening effect in the 6082 alloy. The lengths of the precipitates range from 10 nm to 20 nm, which possibly provides the pinning effect.

According to the orientation relationship between the β'' precipitates and the Al matrix as $\langle 100 \rangle_{\beta''} // \langle 100 \rangle_{\text{Al}}$, some of the two kinds of diffraction spots overlap completely, which means the β'' phase is coherent with the Al matrix [17]. As shown by the local magnification in Figure 6a, the bright-field image of the β'' precipitates presents a semilunar shape with

two different dark contrasts and no contrast in the intermediate position. This further confirms that the β'' phase is coherent with the Al matrix.

Needle-shaped β'' precipitates (about 20 nm in length) and rod-shaped β' precipitates (about 50–60 nm in length) can be observed. Both precipitates are small and occur with a high density. The β' precipitates are the stabilizing product of the β'' precipitates and are semi-coherent with the matrix, which concurs with the only slightly decreased hardness.

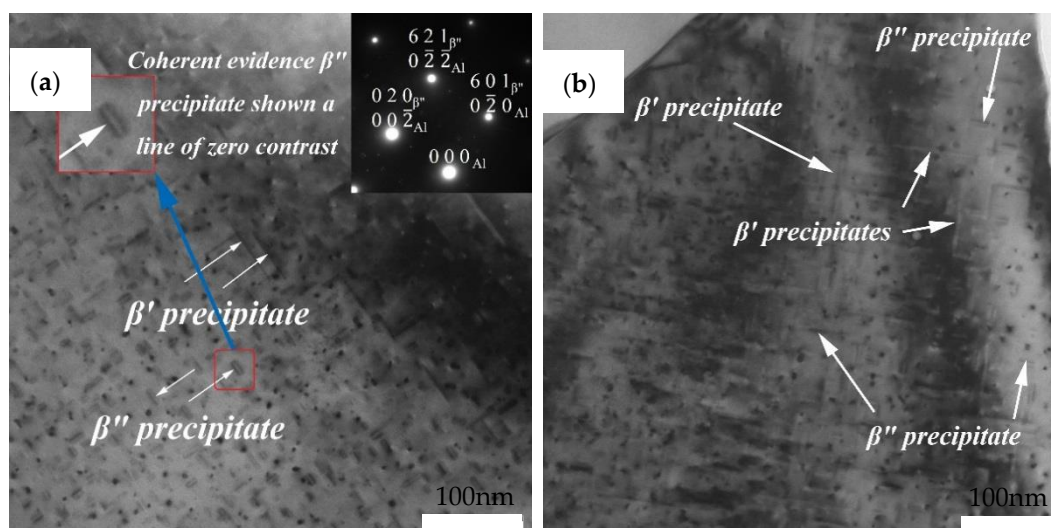


Figure 6. Typical TEM images of 6082 alloy aged for 6 h at 170 °C (a) and 190 °C (b).

From the longitudinal section of the particles, it can be seen that these precipitates are all vertically distributed. The round granular particles are the cross-sections of the precipitates. Compared with the 170 °C specimen (Figure 6a), there is evidence that the 190 °C (Figure 6b) sample contains more rod-shaped β' precipitates and less of the β'' metastable phase, according to the cross-section of these two precipitates. This suggests that the very fine structure of the β'' precipitates is related to the maximum hardness and tensile strength. The β'' precipitate is coherent with the matrix, offering a strong pinning effect during deformation. Meanwhile, the occurrence of more and more semi-coherent β' precipitates substituting the coherent β'' precipitates causes a continuous reduction in local distortion, leading to a corresponding decrease in strength and hardness. The precipitate transition results in a variation in TS and hardness between 170 °C and 190 °C.

4. Discussion

This section focuses on the two TEM-anchored states specific to the present 6082 batch and their direct link to their properties. At 170 °C/6 h, BF-TEM near $\langle 100 \rangle_{Al}$ shows needle-like β'' (~10–20 nm), with a characteristic double-dark coherency contrast; at 190 °C/6 h, rod-like β' (~50–60 nm) becomes more prevalent with reduced coherency. These two states bracket the strength/hardness peak and the first softening regime. As many previous studies have reported, the main strengthening phase of several precipitates evolved from an initial super-saturated solid solution is formed of needle-like β'' precipitates [18].

At the very early stage, the clustering of solute atoms occurs at the beginning of the aging sequence, which could be detected by the heat release of the reaction with differential scanning calorimetry (DSC). The vacancies introduced in the quenching stage started to dissociate, annealing out through the free surface or grain boundaries and Mg atom compounds with the Si atom [19]. The precursor of the β'' phase, traditionally called the G.P. I zone, has been detected and analyzed, proving that it has a unit cell that is just slightly

different from the needle-like β'' phase [9]. The transformation of the series precipitates could impact the alloy's hardening behavior.

Coherent β'' imposes a strong local distortion field that pins moving dislocations, maximizing TS/HB near 170 °C; as the β' fraction rises at 190 °C, coherency is partially relieved, lowering the obstacle strength and producing the moderate softening we measure, which is shown by the 2 h aging samples (the red age hardening curve in Figure 2). For the 2 h sample, the strength from 90 °C to 140 °C was improved by 25 MPa, while the TS from 140 °C to 170 °C increased to 90 MPa. The difference in the phase interface between the precipitates and the matrix leads to a divergence in strengthening.

As is widely known, β' precipitates are semi-coherent with the Al matrix, and the influence of the coherency strain from β' to the matrix is less than that of β'' [20]. This could explain the decrease in the strength of the 190 °C sample compared to that of the 170 °C one (Figure 2). The TEM image also shows that more rod-shaped β' precipitates were observed in the 190 °C sample image, while more needle-like β'' precipitates occurred in the 170 °C sample. In addition, the 140 °C/170 °C/190 °C age hardening curves (Figure 4b) reveal that a higher aging temperature could move the peak value stage forward. This implies that the precipitate transition time was reduced. A higher diffusion activation energy plays a crucial role in the diffusion of Mg and Si, which is necessary for the transformation of the precipitates.

Aluminum is a face-centered cubic crystal, while β'' precipitates have been reported by several scholars as base-centered monoclinic crystals [7,8,11]. The precipitates are different from the matrix in terms of the structure and lattice constants, leading to an inhomogeneous distorted region and coherent strain field formation, which pins moving dislocations and elevates the cutting stress; this is the primary origin of the strength/hardness peak. As the fraction of semi-coherent β' increases, the coherency strain is relieved, and the effective obstacle strength decreases, producing a modest drop in TS/HB while permitting a rebound in ductility (TE). This micro-to-macro linkage is consistent with the strain-field schematic (interaction of neighboring precipitates) and explains the property changes between 170 °C and 190 °C. The aging curves demonstrate that an increase in temperature advances the time to peak and shortens the interval for the $\beta'' \rightarrow \beta'$ transition, indicating kinetic acceleration of solute diffusion and transformation. Within the present time window, this means that 170 °C favors a β'' -dominated population (peak strengthening), whereas 190 °C reaches the early over-aging regime, in which β' becomes more prevalent and softening begins. The cross-section of the precipitates is given in Figure 7a, which presents the strain field around the strengthening precipitate particles. The fully coherent structure with low misfit is illustrated in Figure 7b. Strain contrast occurs at the Al matrix planes that are distorted by the β'' precipitates [17]. The strain field caused by the two neighboring particles could affect each other (as shown in Figure 7c). During deformation, the dislocation movement was mainly hindered by the pinning effect of the β'' precipitates, which is responsible for the increase in tensile strength.

Grounded in this mechanism, β'' dominates, and extensive β' has not yet compromised strength. Figure 8 shows the relationship of time–temperature–property (TTP) and reflects the optimal aging window for the 6082 aluminum alloy. The 160–180 °C range for the 2–4 h window reliably balances TS $\approx$ 360 MPa, HB $\approx$ 105, and TE > 10% in this batch.

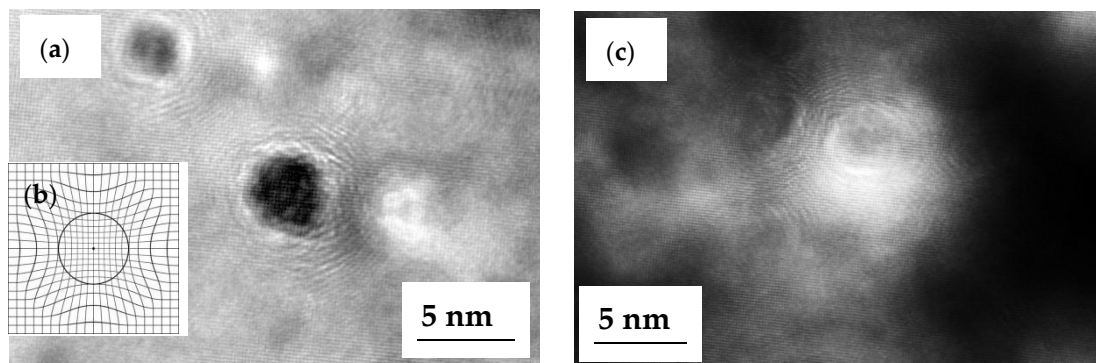


Figure 7. The cross-section of the β'' precipitation fully coherent to the matrix with low misfit (a), sketch (b) and the interaction of the neighboring precipitated particles (c).

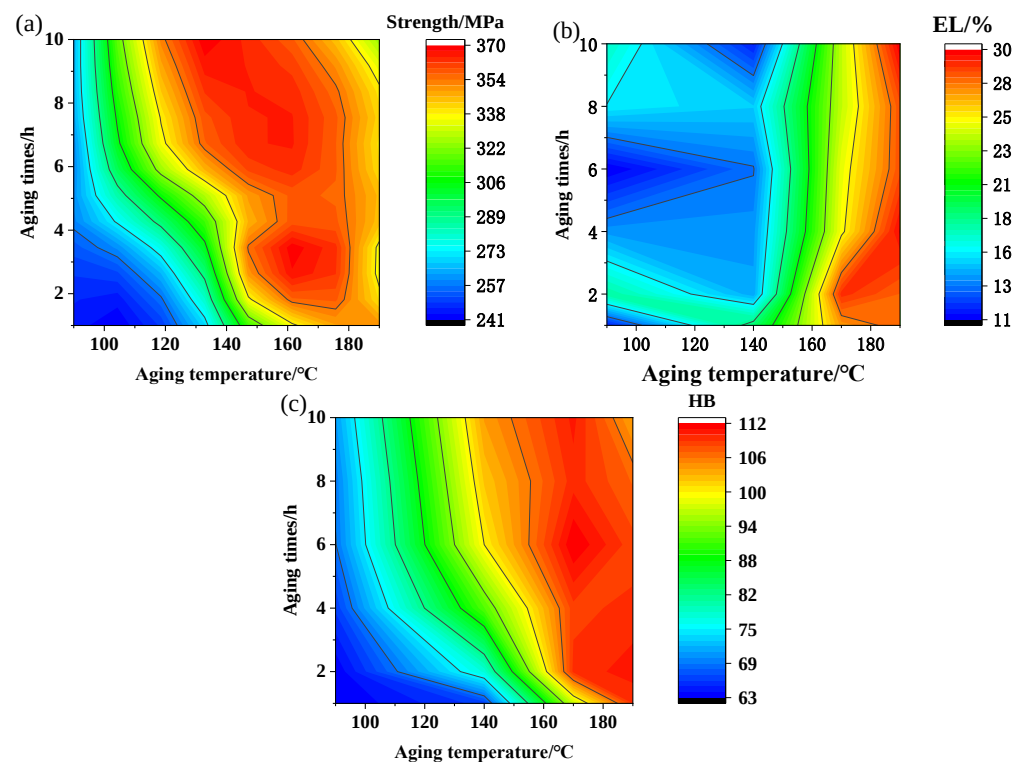


Figure 8. TTP diagram: (a) strength; (b) elongation; (c) hardness.

5. Conclusions

A controlled artificial aging treatment over a wide temperature and time range was conducted on a commercial 6082 aluminum alloy. The properties and hardness were subsequently determined. The microstructure and fracture morphology were characterized through microscopic observation. The following conclusions were drawn:

1. The alloy under the economical aging treatment procedure (aged at 160~180 °C for 2~4 h) obtain potential mechanical properties, with a combination of a tensile strength of about 360 MPa, hardness of 105 HB, and total elongation over 10%.
2. The fracture surface of the tensile test specimens of the artificially aged 6082 alloy was ductile. As the aging temperature increased, the size of the fracture surface dimples decreased from 8–10 μm (aging at 90 °C) μm to 4–7 μm (aging at 170 °C).
3. Using transmission electron microscopy, β'' precipitates with a coherent strain field were successfully captured and characterized. The coherent distortion effect induced

by the β'' precipitates produced a strengthening effect, which is the essence of aging strengthening in 6082 alloy.

4. Over-aging causes the transformation of small-sized (20 nm in length) coherent β'' precipitates into large-sized (50~60 nm in length) semi-coherent β' precipitates, resulting in a slight decrease in the material's tensile strength and hardness.

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Conflicts of Interest: The authors declare that they have no known competing financial interests or personal relations that could have appeared to influence the work reported in this paper.

References

1. Ma, W.Y.; Wang, B.Y.; Yang, L.; Tang, X.; Xiao, W.; Zhou, J. Influence of solution heat treatment on mechanical response and fracture behavior of aluminum alloy sheets: An experimental study. *Mater. Des.* **2015**, *88*, 1119–1126. [\[CrossRef\]](#)
2. Marioara, C.D.; Andersen, S.J.; Jansen, J.; Zandbergen, H.W. The influence of temperature and storage time at RT on nucleation of the β'' phase in a 6082 Al–Mg–Si alloy. *Acta Mater.* **2003**, *51*, 789–796. [\[CrossRef\]](#)
3. Takeda, M.; Ohkubo, F.; Shirai, T. Stability of metastable phases and microstructures in the ageing process of Al–Mg–Si ternary alloys. *J. Mater. Sci.* **1998**, *33*, 2385–2390. [\[CrossRef\]](#)
4. Bordeasu, I.; Ghiban, B.; Micu, L.M.; Luca, A.N.; Demian, A.M.; Istrate, D. The influence of heat aging treatments on the cavitation erosion behavior of a type 6082 aluminum alloy. *Materials* **2023**, *16*, 5875. [\[CrossRef\]](#) [\[PubMed\]](#)
5. Murayama, M.; Hono, K. Pre-precipitate clusters and precipitation processes in Al–Mg–Si alloys. *Acta Mater.* **1999**, *47*, 1537–1548. [\[CrossRef\]](#)
6. Pogatscher, S.; Antrekowitsch, H.; Leitner, H.; Ebner, T.; Uggowitzer, P.J. Mechanisms controlling the artificial aging of Al–Mg–Si Alloys. *Acta Mater.* **2011**, *59*, 3352–3363. [\[CrossRef\]](#)
7. Ravia, C.; Wolverton, C. First-principles study of crystal structure and stability of Al–Mg–Si–(Cu) precipitates. *Acta Mater.* **2004**, *52*, 4213–4227. [\[CrossRef\]](#)
8. Andersen, S.J.; Zandbergen, H.W.; Jansen, J.; Trøholt, C.; Tundal, U.; Reiso, O. The crystal structure of the β'' phase in Al–Mg–Si alloys. *Acta Mater.* **1998**, *46*, 3283–3298. [\[CrossRef\]](#)
9. Marioara, C.D.; Andersen, S.J.; Jansen, J.; Zandbergen, H.W. Atomic model for G.P.-zones in a 6082 Al–Mg–Si System. *Acta Mater.* **2001**, *49*, 321–328. [\[CrossRef\]](#)
10. Buha, J.; Lumley, R.N.; Crosky, A.G.; Hono, K. Secondary precipitation in an Al–Mg–Si–Cu alloy. *Acta Mater.* **2007**, *55*, 3015–3024. [\[CrossRef\]](#)
11. Zandbergen, H.W.; Andersen, S.J.; Jansen, J. Structure determination of Mg₅Si₆ particles in Al by dynamic electron diffraction studies. *Science* **1997**, *277*, 1221–1225. [\[CrossRef\]](#)
12. Yang, L.; Bai, D.; Wang, D.; Liu, C.; Yang, S.; Huang, G. Effect of aging temperature on microstructure and corrosion behavior of 6082 aluminum alloy. *Mater. Today Commun.* **2023**, *36*, 106583. [\[CrossRef\]](#)
13. Bartawi, E.H.; Shaban, G.; Ambat, R. Role of aging time and Cu/Zn additions on the microstructure and intergranular corrosion resistance of 6082 Al–Mg–Si alloy. *Corros. Sci.* **2025**, *254*, 113027. [\[CrossRef\]](#)
14. Hu, Z.; Zheng, J.; Pang, Q.; Sun, Q.; Zhao, N. Influence of strain rates and aging time on microstructure and hardness of integrally compressed 6082 aluminum alloy. *J. Mater. Res. Technol.* **2023**, *27*, 826–838. [\[CrossRef\]](#)
15. Jiang, J.; Cui, J.; Wang, Y.; Huang, M.; Dong, J.; Yan, J. Effect of T6 heat treatment on microstructure and mechanical properties of 6082 aluminum alloy automotive flange components formed by squeeze casting. *Mater. Sci. Eng. A* **2024**, *912*, 146974. [\[CrossRef\]](#)
16. GB/T228.1-2010; Metallic Materials—Tensile Testing—Part 1: Method of Test at Room Temperature. Standards Press of China: Beijing, China, 2010.

17. Rong, Y.H. *Introduction to Analytical Electron Microscopy*; Higher Education Press: Beijing, China, 2015; pp. 286–287.
18. Gaber, A.; Afify, N.; Mostafa, M.S.; Abbady, G. Effect of heat treatment on the precipitation in Al-1 at.% Mg-x at.% Si (x = 0.6, 1.0 and 1.6) alloys. *J. Alloys Compd.* **2009**, *477*, 295–300. [[CrossRef](#)]
19. Afify, N.; Gaber, A.; Mostafa, M.S.; Abbady, G. Influence of Si concentration on the precipitation in Al-1 at.% Mg alloy. *J. Alloys Compd.* **2008**, *462*, 80–87. [[CrossRef](#)]
20. Edwards, G.A.; Stiller, K.; Dunlop, G.L.; Couper, M.J. The precipitation sequence in Al-Mg-Si Alloys. *Acta Mater* **1998**, *46*, 3893–3904. [[CrossRef](#)]

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