

Effect of initial tempers on mechanical properties of creep-aged AA2050

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Abstract. The evolution of mechanical properties of a third-generation Al–Cu–Li alloy, AA2050, with different initial tempers (as-quenched WQ, naturally aged T34 and peak-aged T84) during creep-ageing has been investigated in this study. A set of creep-ageing tests was carried out under 150 MPa at 155 °C with different durations for all initial temper conditions and tensile tests were performed subsequently to acquire the main mechanical properties of the creep-aged alloys, including the yield strength, ultimate tensile strength and uniform elongation. The evolution of these mechanical properties during creep-ageing has been discussed in association with precipitation behaviour of AA2050 alloys with different initial tempers. The results indicate that the T34 alloy is the best choice for creep age forming (CAF) applications among these initial tempers, as it provides better yield strength and uniform elongation concurrently after creep-ageing. In addition, a work hardening rate analysis has been carried out for all the creep-aged alloys, helping to understand the detailed dislocation/precipitate interaction mechanisms during plastic deformation in the creep-aged AA2050 alloys with WQ, T34 and T84 initial tempers.

Keywords: Initial temper / mechanical property / Al–Cu–Li alloy / AA2050 / creep age forming / work hardening

1 Introduction

Creep age forming (CAF) is an advanced forming technology originally proposed to manufacture large/extra-large panels in the aerospace industry [1]. Creep and age hardening occur simultaneously to concurrently deform and strengthen aluminium alloys when load is applied to the alloys at a high temperature during CAF [2,3]. The recently developed third-generation Al–Cu–Li alloys have been reported as a promising material for aerospace applications due to their high stiffness and strength-to-weight ratio advantages [4,5]. The creep-ageing behaviour of a third-generation Al–Cu–Li alloy AA2050 has been investigated in previous studies [6,7], which built a good foundation for the alloy to be manufactured through CAF process. For potential industrial applications, the mechanical properties of the alloy after CAF are important, especially for the aerospace components, and therefore need to be investigated and evaluated.

It has been widely reported that initial tempers could highly affect the precipitation progress of aluminium alloys during artificial ageing and result in different yield strengths [8]. Different precipitation behaviour has been found during artificial ageing of aluminium alloys, such as Al–Mg–Si [9] and Al–Li [10,11] alloys, with as-quenched or naturally aged initial tempers. An initial drop of yield strength has been particularly observed for alloys with naturally aged initial temper due to the dissolution of pre-existing precipitates or clusters at the early stage of artificial ageing [9]. For Al–Zn–Mg alloy, a pre-ageing step is generally carried out before final artificial ageing, in order to facilitate the homogeneously precipitation and achieve high yield strength [12]. For Al–Cu–Li (Al–Li) alloys, it has been reported that initial tempers with or without pre-stretch will lead to apparent different precipitation and yield strength evolutions during artificial ageing [11,13]. Kumar et al. [11] indicated that pre-stretch can assist the nucleation of T_1 (Al_2CuLi) precipitates and accelerate the age hardening of the alloy, enhancing the yield strength of the naturally aged Al–Cu–Li alloys. Similar phenomenon has also been observed in the as-quenched Al–Cu–Li alloys [13,14].

For CAF process, the creep-ageing behaviour of an Al–Cu–Li alloy (AA2050) with different initial tempers,

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Table 1. Main chemical composition of AA2050 (wt%).

Al	Cu	Li	Mg	Ag	Mn	Zr
Balance	3.6	0.9	0.34	0.35	0.34	0.08

namely the as-quenched (WQ), naturally aged T34 and peak-aged T84, has been investigated, and different creep and age hardening behaviour have been reported for different initial tempers under the same stress level at 155 °C [6]. The microstructures and yield strength of an as-quenched Al–Li alloy after CAF process have also been investigated and the results indicated an enhanced yield strength in the formed material due to the applied stress [15]. However, only the yield strength of the alloys after creep-ageing was considered in previous studies, other mechanical properties, such as ultimate tensile strength (UTS) and uniform elongation, are also important for the CAFed alloys and should be considered as well, so as to provide a comprehensive method to select the proper initial temper condition for alloys for CAF applications.

Meanwhile, the work hardening behaviour of Al–Cu–Li alloys also attracts much interest, as it reflects the tensile strength and elongation of the alloys and can act as a tool to indicate the detailed dislocation/precipitate interaction mechanism during plastic deformation. Earlier studies reported that Al–Cu–Li alloys with different precipitates show different work hardening behaviour during plastic deformation, depending on the main precipitates in the material (such as T_1 and/or δ' (Al_3Li)) [16–18]. As different precipitates exist after creep-ageing in the alloys with different initial tempers, their effects on the plastic deformation of creep-aged alloys are also of interest.

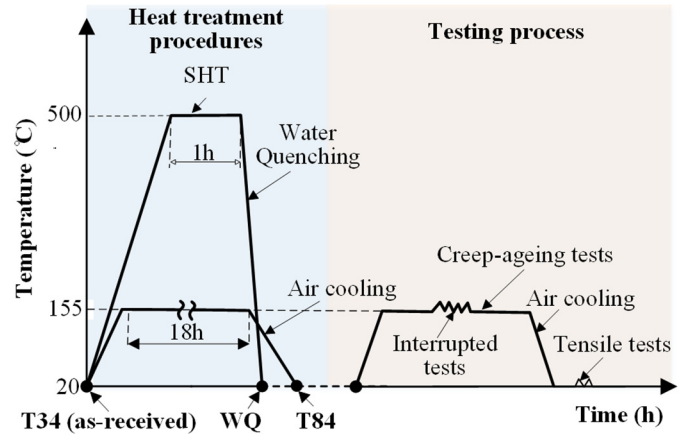
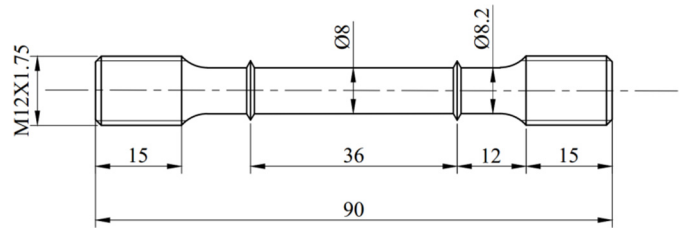
In this paper, an Al–Cu–Li alloy, AA2050, with three commonly reported initial temper conditions, as-quenched, naturally aged and peak-aged, was studied to investigate the effect of initial temper on evolutions of mechanical properties, including yield strength, UTS and uniform elongation, during creep-ageing in CAF process. The plastic deformation behaviour of the creep-aged alloys is then discussed based on the work hardening rate analysis, so as to understand the detailed dislocation/precipitate interaction mechanisms in creep-aged AA2050 alloys.

2 Material and experiments

2.1 Material and heat treatment

The material used in this study is a third-generation Al–Cu–Li alloy AA2050, provided by Embraer (Brazil). The main chemical composition of the alloy is listed in Table 1. The as-received material was 12.7 mm thick plate and at T34 temper, which had been solution heat treated (SHT), water quenched, pre-stretched and naturally aged.

Three commonly used initial temper conditions, namely as-quenched (WQ), naturally aged (T34) and peak-aged (T84), were selected for investigation in the current study. The materials with WQ and T84 initial tempers were prepared by heat treatment of the as-received T34 material. WQ initial temper was prepared by solution

**Fig. 1.** Schematic of heat treatment procedures and subsequent testing conditions.**Fig. 2.** Specimen geometry and dimensions for creep-ageing and tensile tests (units: mm).**Table 2.** Experimental plan for creep-ageing and tensile tests.

Initial temper	Creep-ageing time (h)
T34	0, 2, 5, 8, 12, 18
WQ	0, 5, 12, 18, 24
T84	0, 6, 15, 50

heat treating (SHT) at 500 °C for 1 h, followed by water quenching, while T84 initial temper was obtained by artificial ageing of the as-received material at 155 °C for 18 h. Figure 1 schematically shows the heat treatment procedures for materials preparation and subsequent creep-ageing and tensile tests in this study.

2.2 Creep-ageing and tensile tests

The same testing specimens were used for creep-ageing and subsequent room temperature tensile tests, whose dimensions are shown in Figure 2. The specimens were machined from an as-received thick plate along the rolling direction. The heat treatment procedures indicated in Section 2.1 were then performed to prepare specimens with WQ and T84 initial tempers.

The prepared AA2050 specimens with WQ, T34 and T84 initial tempers were creep-aged under a constant stress level of 150 MPa at 155 °C for different durations, as listed in Table 2. For the alloy with WQ initial temper,

Table 3. Summary of precipitation progress during artificial-ageing/creep-ageing of AA2050 with WQ, T34 and T84 initial tempers at 155 °C (summarised from [6,13,16,19]).

Initial temper	Initial state	Precipitation progress	Final state
WQ	SSSS	Nucleation and growth of GP zones and δ' → Nucleation and growth of T_1 and minor θ'/S'	T_1 and δ' with minor θ'/S' (24 h)
T34	Cu-rich clusters	Dissolution of Cu clusters → Nucleation and growth of precipitates (mainly T_1 with minor θ')	T_1 with minor θ' (18 h)
T84	T_1 with minor θ'	Coarsening of T_1 (slight increase in precipitate length and volume fraction)	T_1 with minor θ' (50 h)

creep-ageing tests were performed immediately after water-quenching. Subsequent tensile tests were carried out at room temperature to obtain the main mechanical properties of the creep-aged alloys. The heat treatment conditions for these tests are shown in Figure 1.

Both creep-ageing and tensile tests were performed on an Instron 5584 machine. A high temperature extensometer Instron 2620-601 was attached onto the specimen to capture the strain during both tests. For the creep-ageing tests, an assisting furnace with an accuracy of $\pm 3^\circ\text{C}$ was attached to the Instron machine to provide a high temperature testing environment. The constant stress was applied after the specimens reached 155 °C. For subsequent tensile tests, a strain rate of 10^{-4} s^{-1} was used.

3 Evolution of precipitates

The details of precipitation progress of AA2050 and other similar Al–Cu–Li alloys artificially aged/creep-aged from WQ or naturally aged (T34) initial tempers to over-aged states have been investigated and discussed before [6,13,16,19]. A brief summary of the precipitation progress is given in this section to assist the further analysis of the current study. As listed in Table 3, different microstructures exist in the Al–Cu–Li alloy AA2050 with initial tempers of WQ, T34 and T84. The WQ alloy is composed of a supersaturated solid solution (SSSS), the naturally aged T34 alloy has Cu-rich clusters; while in the peak-aged T84 alloy, T_1 precipitates with minor θ' (Al_2Cu) play the dominant role. Different precipitation progress during artificial ageing is summarised below for these alloys:

- For WQ alloy, GP zones and δ' appear first, followed by the nucleation and growth of T_1 precipitates (with minor θ' and S' (Al_2CuMg)). After 24 h creep-ageing of the alloy at 155 °C, both T_1 and δ' precipitates (with minor θ' and S') are expected in the alloy [6,11].
- T34 alloy experiences an initial dissolution of Cu-rich clusters at the first 2 h of creep-ageing at 155 °C and an accelerated nucleation of T_1 precipitates (with minor θ') subsequently due to the dislocations in the alloys introduced by pre-stretching [20]. After 18 h creep-ageing at 155 °C, T34 alloy reaches the peak-ageing state, with mainly T_1 precipitates [6,16], which can be seen in the transmission electron microscopy (TEM) image

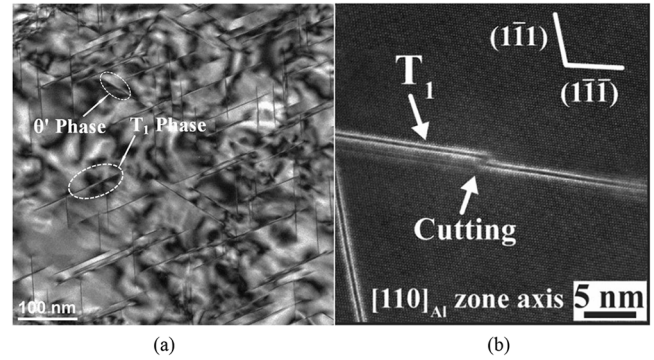


Fig. 3. Precipitates in creep-aged T34 alloy under 150 MPa at 155 °C for 18 h: (a) TEM image showing the T_1 and θ' precipitate [6] and (b) HR-TEM image showing shearing (cutting) of the single layer T_1 precipitate [21].

shown in Figure 3a [6]. The T_1 precipitates in the alloy at the peak-aged state have been reported as having a single layer property whose dominant strengthening mechanism is shearing, as illustrated by the high-resolution TEM (HR-TEM) image in Figure 3b [21].

- For T84 alloy, it has been reported that the single layer T_1 precipitates would play the dominant role in further artificial ageing at 155 °C for a long time after peak-ageing [22]. T_1 precipitates experience a slight growth with a minor increase in length, while the thickness remains constant for a long ageing time, up to 200 h [19]. A slight increase in volume fraction of T_1 precipitates was also observed at the first couple of hours [19,23].

4 Results and discussion

Test results for the yield strength, UTS and uniform elongation of the creep-aged AA2050 alloys with WQ, T34 and T84 initial tempers will be presented and discussed in Section 4.1. Work hardening analysis of stress-strain curves of all the creep-aged alloys will be performed in Section 4.2 to understand the detailed dislocation/precipitate interaction mechanisms during plastic deformation.

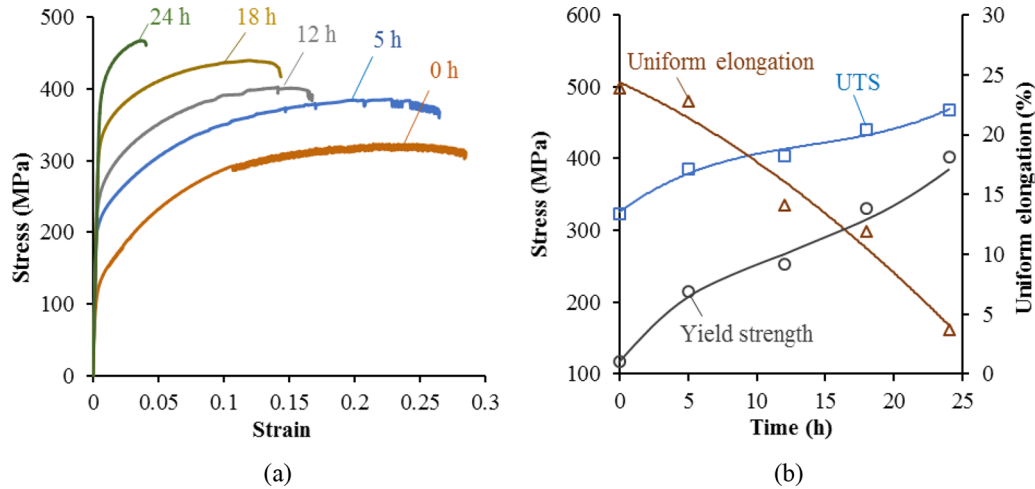


Fig. 4. (a) Tensile stress–strain curves of AA2050-WQ creep-aged under 150 MPa at 155 °C with indicated durations and (b) evolutions of yield strength, UTS and uniform elongation of the creep-aged WQ alloy.

4.1 Effect of initial tempers on mechanical properties

4.1.1 WQ initial temper

Figure 4 shows the tensile test results of the AA2050-WQ alloy creep-aged under 150 MPa at 155 °C for different time. The initial WQ alloy is in SSSS state and exhibits a comparative low yield strength of 117 MPa. A substantial work hardening behaviour is observed in the alloy during plastic deformation, which can be explained by the high efficiency of dislocation storage during plastic deformation due to the high level of free solutes in the SSSS of the alloy [17]. In addition, significant serrations, known as the Portevin–Le Chatelier effect, are observed in the stress–strain curve of the initial WQ alloy, which is also related to the SSSS state of the alloy, as sufficient free solutes in the alloy matrix can diffuse around dislocations to restrain its movement, leading to serrations during plastic deformation [17,24].

The yield strength increases continuously with the creep-ageing time for AA2050-WQ alloy because of the nucleation and growth of the precipitates summarised in Table 3. In the meantime, in the early stage of creep-ageing, within 12 h at 155 °C in this study (Fig. 4a), serrations still exist in the stress–strain curves of the creep-aged WQ alloy. Deschamps et al. [17] observed a similar serration phenomenon during artificial ageing of AA2196 at the stage where δ' precipitates are the dominating factor on yield strength evolution and indicated that the phenomenon is related to the properties of δ' precipitates. After 12 h creep-ageing, yield strength increases at a higher rate and serrations disappear in the stress–strain curves, which is believed to be caused by the beginning of nucleation and growth of T_1 precipitates [17]. The alloy is still in under-ageing condition after 24 h creep-ageing in the current study and the yield strength reaches 402 MPa.

Figure 4b illustrates the evolutions of yield strength, UTS and uniform elongation during creep-ageing of WQ alloy. The UTS data shares the similar trend to yield strength, increasing continuously within 24 h. Meanwhile, the level of work hardening strength (difference

between UTS and yield strength) decreases with increasing creep-ageing time and a much higher decreasing rate can be observed after 12 h, when T_1 precipitates start to nucleate as mentioned above. Similar to the work hardening level, the uniform elongation of the creep-aged WQ alloy decreases continuously with the creep-ageing time, which is also related to the free solute contents in the alloy [23].

4.1.2 T34 initial temper

Figure 5 shows the tensile data for various creep-aged AA2050-T34 alloy. According to the stress–strain curves in Figure 5a, the as-received T34 alloy shows a relatively high yield strength of about 269 MPa with a significant work hardening behaviour. Apparent serrations can also be observed in the stress–strain curve of the as-received T34 alloy, which indicate that only minor Cu-rich clusters exist in the as-received alloy and sufficient solutes still remain in the alloy matrix to restrain dislocation movement [24,25].

After 2 h creep-ageing, the yield strength of the alloy drops to about 245 MPa, which is due to the dissolution of Cu-rich clusters in the T34 alloy. After that, yield strength increases with the creep-ageing time and nearly recovers to the as-received level (269 MPa) at about 5 h. Significant serrations are also observed in the stress–strain curve of the 2 h creep-aged T34 alloy, reflecting the dissolution of clusters in the first 2 h of creep-ageing. For the 5 h creep-aged T34 alloy, only slight serrations can be found, indicating that nucleation and growth of precipitates begin and less free diffused solutes exist in the matrix [25]. All these results correspond well with the precipitation progress of T34 alloy summarised in Table 3.

After 5 h creep-ageing, accelerated nucleation and growth of T_1 precipitates occur, resulting in the dramatical increase of yield strength until peak-ageing state is achieved after 18 h, with a maximum yield strength of about 498 MPa. In addition, serrations in stress–strain curves disappear with the presence of T_1 precipitates after 5 h creep-ageing.

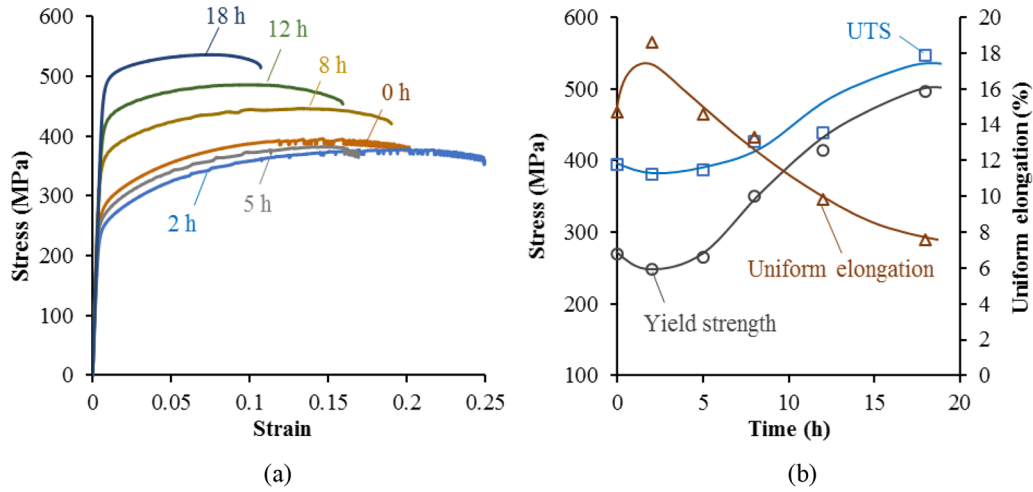


Fig. 5. (a) Stress–strain curves for AA2050-T34 alloy creep-aged under 150 MPa at 155 °C with indicated durations and (b) evolutions of yield strength, UTS and uniform elongation of the creep-aged T34 alloy.

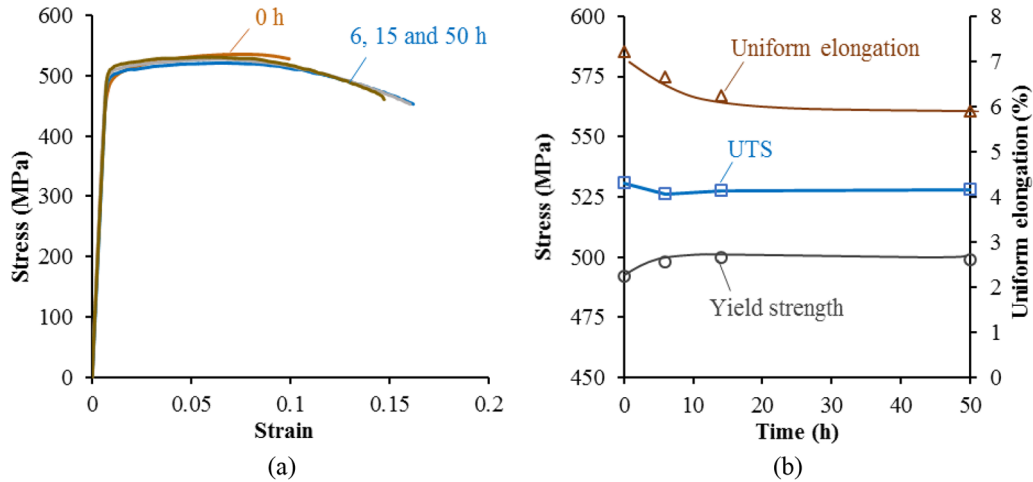


Fig. 6. (a) Tensile curves of AA2050-T84 alloy creep-aged under 150 MPa at 155 °C for different durations and (b) evolutions of yield strength, UTS and uniform elongation of the creep-aged T84 alloy.

Figure 5b demonstrates the evolutions of yield strength, UTS and uniform elongation during creep-ageing of T34 alloy. UTS shares the similar pattern to yield strength along the creep-ageing time, dropping down in the first 2 h and increasing substantially up to 18 h. However, the difference between UTS and yield strength becomes smaller with increasing creep-ageing time, indicating a weakened work hardening behaviour due to the well-developed precipitates [17]. Meanwhile, the uniform elongation evolves oppositely to yield strength and UTS, increasing at the first 2 h and decreasing afterwards.

4.1.3 T84 initial temper

Figure 6a shows the stress–strain curves of AA2050-T84 alloy after creep-ageing. A similar stress–strain relationship is observed in all creep-aged alloys, from 0 to 50 h creep-ageing investigated in this study. The same

behaviour has been reported by Deschamps et al. [17] during further artificial ageing of another peak-aged Al–Cu–Li alloy AA2198. The stress–strain results correspond well with the precipitation behaviour of the T84 alloy summarised in Table 3, as the T_1 precipitates remain in the alloy for a long time with only a slight growth in precipitate length and volume fraction during artificial ageing at 155 °C (up to 200 h after peak-ageing [19]). Meanwhile, a minor work hardening behaviour is observed in all creep-aged T84 alloys, due to the severe depletion of free solutes by the well-developed T_1 precipitates.

The evolutions of yield strength, UTS and uniform elongation of the creep-aged T84 alloy are summarised in Figure 6b. A slight increase of yield strength at the first few hours is observed, which may be caused by the minor growth of T_1 precipitates at the beginning of artificial ageing, as presented in Table 3. After that, the high yield strength remains in the alloy for a long time, up to 50 h in

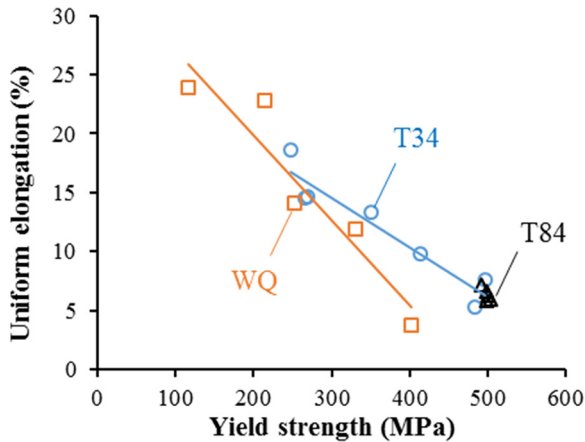


Fig. 7. Relationship between uniform elongation and yield strength for the creep-aged AA2050 alloys with WQ, T34 and T84 initial tempers.

this study. This phenomenon has been explained by the remained single layer property of T_1 precipitates even after a long-term artificial ageing at 155 °C, with which shearing is still the dominant strengthening mechanism of the alloy [22,23]. On the contrary, the uniform elongation experiences an initial decrease due to some further depletion of solid solutes for precipitation, and then stays at a stable level when the equilibrium state of solutes is achieved, as demonstrated in Figure 6b.

4.1.4 Comparison

As CAF is mainly used for manufacturing structural panel products, both high strength and high ductility are preferred for the creep-aged material. The tensile data in Figures 4 and 5 show that for the same creep-ageing time, AA2050-T34 alloy has a higher yield strength but a lower uniform elongation compared to the WQ alloy. While after peak-ageing (T84), the alloy has high yield strength and low uniform elongation for a long time in subsequent creep-ageing at 155 °C.

To evaluate the performance of creep-aged materials, the uniform elongation is plotted against the yield strength in Figure 7 for the creep-aged WQ, T34 and T84 alloys. An inverse relationship can be seen between the yield strength and uniform elongation for WQ and T34 alloys. An interesting result is observed that with the same yield strength level, the creep-aged T34 alloy has a comparatively higher uniform elongation than the creep-aged WQ alloy, indicating that AA2050 alloy with T34 initial temper could be a better choice for CAF process than WQ alloy, as both comparatively high yield strength and uniform elongation can be achieved concurrently after creep-ageing of T34 alloy. This phenomenon is believed to be related to the different precipitates in the creep-aged T34 and WQ alloys. As T_1 precipitates contribute to a higher hardening effect and nucleate more homogeneously than δ' precipitates [26], better mechanical properties can be expected for the creep-aged T34 alloy that contains mainly T_1 precipitates, than WQ alloy, in which both T_1 and δ' exist. Further creep-ageing at 155 °C for up to 50 h contributes nearly

little effect on the strength and elongation evolution for T84 alloy; however, lower creep strain has been reported in T84 alloy than that in T34 alloy under the same stress level during creep-ageing [6]. Hence, it is reasonable to conclude that T34 initial temper is the best choice for AA2050 alloy among the three initial temper conditions investigated in this study for the CAF process applications.

4.2 Work hardening analysis

During tensile tests of the creep-aged alloys, the interaction between dislocations and precipitates determines the plastic deformation behaviour of the alloys. Analysis is carried out in this section on the work hardening behaviour to reveal possible dislocation/precipitate interaction mechanisms during plastic deformation of the creep-aged AA2050 alloys with different initial tempers. Kocks–Mecking plots [27] are employed for the work hardening analysis and are shown in Figure 8, illustrating the evolution of work hardening rate ($d\sigma/d\varepsilon$, where σ is the applied stress and ε is the strain) against the work hardened strength ($\sigma - \sigma_y$, where σ_y is the yield strength) for all creep-aged AA2050 alloys.

The creep-aged WQ, T34 and T84 alloys generally show similar work hardening behaviour, where the work hardening rate decreases with increasing hardened strength levels. High initial work hardening rates (6000–8000 MPa) are observed in the WQ and T34 alloys, as shown in Figure 8a and b, since high free solute contents exist in the alloy matrix with both initial tempers [17]. A similar work hardening rate is observed for T34 alloys at the early stages of creep-ageing (within 5 h in this study) when mainly Cu-rich clusters or weak precipitates (such as GP zones) exist, as shown in Figure 8b.

The work hardening rate behaviour starts to change when the main strengthening phases (T_1 and/or δ') start to substantially nucleate in the alloys. For T34 alloy, a lower initial work hardening rate with a higher decreasing speed at the end part of the curves in Figure 8b is observed after 5 h creep-ageing. The same work hardening rate behaviour can be observed in the creep-aged WQ alloy after 12 h creep-ageing in Figure 8a. Moreover, the decreasing speed of work hardening rate of the alloys increases with longer creep-ageing time. Similar work hardening rate behaviour with ageing time has been reported in some Al–Li [17] and Al–Zn–Mg [28] alloys with shearable precipitates (such as δ' and η') and has been explained as the occurrence of shearing of precipitates during plastic deformation in tensile tests. It has been reported that T_1 precipitate is a semi-shearable precipitate, which exhibits a single-time shearing property [17]. Hence, shearing would occur in creep-aged T34 alloy with dominant T_1 precipitates (after 5 h creep-ageing) during tensile tests, as demonstrated in Figure 3b, leading to a lower initial work hardening rate and a higher dropping speed at the end of the Kocks–Mecking curves with increasing ageing time [8]. Meanwhile, for the creep-aged WQ alloy, in which both T_1 and δ' exist, a much higher dropping speed of the work hardening rate is observed as δ' has a higher shearable property than T_1 , such as the work hardening rate curve of the 24 h creep-aged WQ alloy shown in Figure 8a.

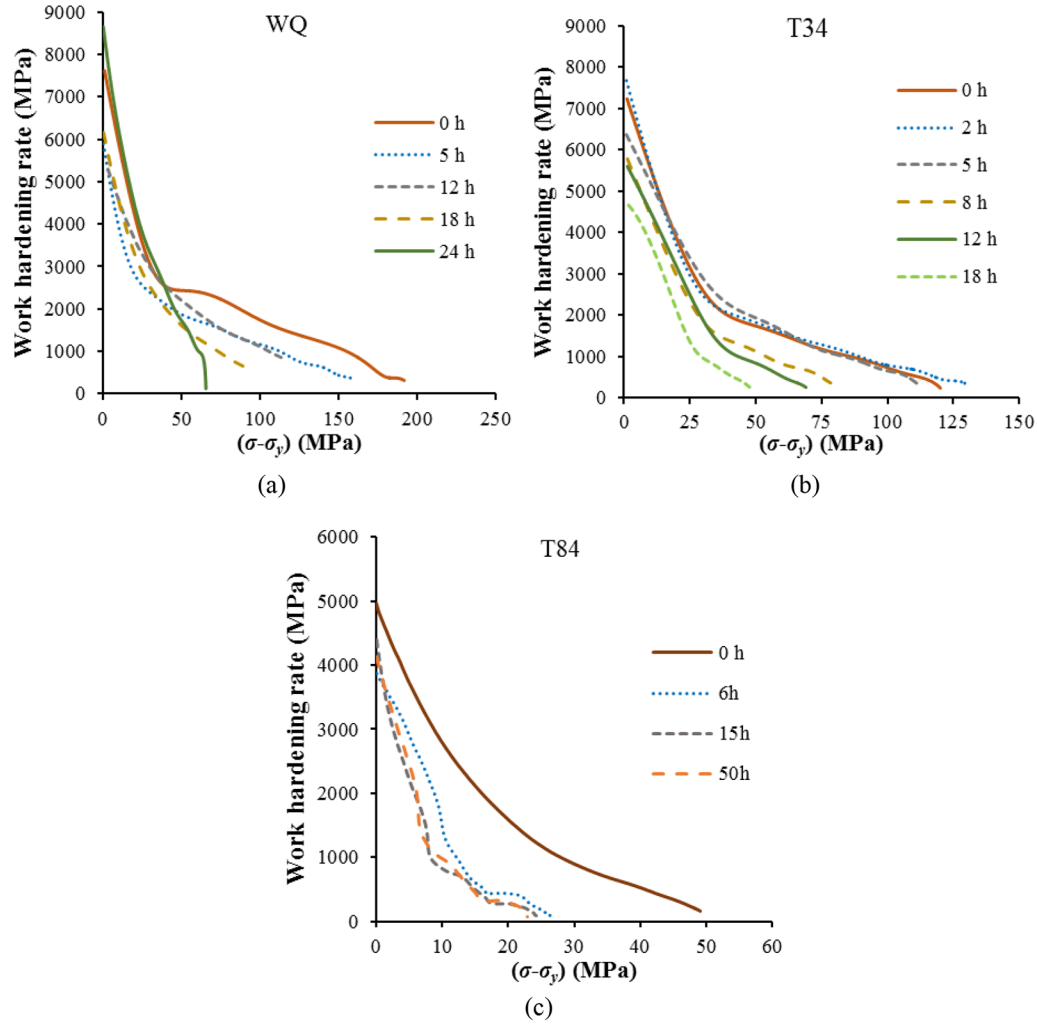


Fig. 8. Evolutions of work hardening rates with the hardening strength $(\sigma - \sigma_y)$ for AA2050 with (a) WQ, (b) T34 and (c) T84 initial tempers creep-aged under 150 MPa at 155 °C for indicated time.

For the T84 alloy, a low initial work hardening rate is observed in Figure 8c for all creep-aged alloys, as T_1 precipitates have already been well-developed in the initial material. The work hardening rate curves for the T84 alloy creep-aged from 6 to 50 h show a similar trend, indicating that similar precipitates exist in these alloys and strengthening of the alloy is still controlled by the shearing mechanism after 50 h creep-ageing at 155 °C.

5 Conclusions

In this study, main mechanical properties of AA2050 with WQ, T34 and T84 initial tempers, creep-aged under 150 MPa at 155 °C for various durations, have been investigated and a work hardening analysis has been carried out to understand the dislocation/precipitate interaction mechanisms during plastic deformation of creep-aged alloys. The following conclusions can be drawn:

(1) Yield strength of AA2050-T34 alloy drops from 269 to 245 MPa within the first 2 h creep-ageing due to dissolution of Cu-clusters and increases substantially

afterward until reaching the peak-ageing state at 18 h. For AA2050-WQ alloy, yield strength increases with creep-ageing time up to the investigated range of 24 h. UTS shares the same trend with yield strength while uniform elongation demonstrates an exactly opposite trend for both creep-aged WQ and T34 alloys. Yield strength, UTS and uniform elongation properties remain at a stable level for AA2050-T84 alloy after creep-ageing for up to 50 h.

- (2) T34 alloy is the best choice for CAF process applications among the three commonly used tempers, as the creep-aged AA2050-T34 alloy can achieve a higher uniform elongation than the creep-aged WQ alloy with the same yield strength and generate higher creep strain levels than T84 alloy at the same stress level.
- (3) The different mechanical properties and work hardening behaviour of creep-aged AA2050 alloys with WQ, T34 and T84 initial tempers are due to their different precipitation behaviour. Creep-aged T34 and T84 alloys contain mainly T_1 precipitate, resulting in a higher strength and a slight lower decreasing speed of work hardening rate than the creep-aged WQ alloy, which contains T_1 and δ' precipitates concurrently.

6 Implications and influences

This work reports the evolution of main mechanical properties of AA2050 with different initial tempers during creep-ageing, providing comprehensive information to select the most appropriate initial temper condition for alloys for CAF applications.

The relationship between microstructures and main mechanical properties, including yield strength, ultimate tensile strength and uniform elongation, has been discussed in detail, providing a solid foundation for future work of unified material modelling, with which evolutions of main mechanical properties during creep-ageing can be predicted to assist CAF process design and optimisation.

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References

1. L. Zhan, J. Lin, T. Dean, *Int. J. Mach. Tool. Manuf.* **51** (2011) 1–17
2. A.C. Lam, Z. Shi, X. Huang, Y.-L. Yang, Y. Zeng, J. Lin, *Manuf. Rev.* **2** (2015) 19
3. K. Ho, J. Lin, T. Dean, *Int. J. Plasticity* **20** (2004) 733–751
4. P. Lequeu, K. Smith, A. Daniélou, *J. Mater. Eng. Perform.* **19** (2010) 841–847
5. H. Gao, T. Weng, J. Liu, C. Li, Z. Li, L. Wang, *Manuf. Rev.* **3** (2016) 9
6. Y. Li, Z. Shi, J. Lin, Y.-L. Yang, B.-M. Huang, T.-F. Chung, J.-R. Yang, *Mater. Sci. Eng. A* **657** (2016) 299–308
7. Y. Li, Z. Shi, J. Lin, Y.-L. Yang, Q. Rong, B.-M. Huang, T.-F. Chung, C.-S. Tsao, J.-R. Yang, D.S. Balint, *Int. J. Plasticity* **89** (2017) 130–149
8. R. Yoshimura, T.J. Konno, E. Abe, K. Hiraga, *Acta Mater.* **51** (2003) 4251–4266
9. W. Poole, D. Lloyd, J. Embury, *Mater. Sci. Eng. A* **234** (1997) 306–309
10. A. Luo, D. Lloyd, A. Gupta, W. Youdelis, *Acta Metall. Mater.* **41** (1993) 769–776
11. K. Kumar, S. Brown, J.R. Pickens, *Acta Mater.* **44** (1996) 1899–1915
12. M. Chemingui, M. Khitouni, K. Jozwiak, G. Mesmacque, A. Kolsi, *Mater. Des.* **31** (2010) 3134–3139
13. S. Ringer, B. Muddle, I. Polmear, *Metall. Mater. Trans. A* **26** (1995) 1659–1671
14. B. Gable, A. Zhu, A. Csontos, E. Starke, Jr., *J. Light Met.* **1** (2001) 1–14
15. H. Li, W. Kang, X. Lu, *J. Alloys Compd.* **640** (2015) 210–218
16. B. Decreus, A. Deschamps, F. De Geuser, P. Donnadieu, C. Sigli, M. Weyland, *Acta Mater.* **61** (2013) 2207–2218
17. A. Deschamps, B. Decreus, F. De Geuser, T. Dorin, M. Weyland, *Acta Mater.* **61** (2013) 4010–4021
18. A.A. Csontos, E.A. Starke, *Int. J. Plasticity* **21** (2005) 1097–1118
19. Y. Yan, L. Peguet, O. Gharbi, A. Deschamps, C. Hutchinson, S. Kairy, N. Birbilis, *Materialia* **1** (2018) 25–36
20. L.-W. Quan, Z. Gang, G. Sam, C. Barry, *Trans. Nonferrous Met. Soc. China* **21** (2011) 1957–1962
21. T.-F. Chung, Y.-L. Yang, C.-N. Hsiao, W.-C. Li, B.-M. Huang, C.-S. Tsao, Z. Shi, J. Lin, P.E. Fischione, T. Ohmura, *Int. J. Lightweight Mater. Manuf.* **1** (2018) 142–156
22. T. Dorin, F. De Geuser, W. Lefebvre, C. Sigli, A. Deschamps, *Mater. Sci. Eng. A* **605** (2014) 119–126
23. T. Dorin, A. Deschamps, F. De Geuser, C. Sigli, *Acta Mater.* **75** (2014) 134–146
24. D. Thevenet, M. Mliha-Touati, A. Zeghloul, *Mater. Sci. Eng. A* **266** (1999) 175–182
25. P. McCormick, *Acta Metall.* **19** (1971) 463–471
26. W. Cassada, G. Shiflet, E. Starke, *Metall. Trans. A* **22** (1991) 299–306
27. U. Kocks, H. Mecking, *Prog. Mater. Sci.* **48** (2003) 171–273
28. G. Fribourg, Y. Bréchet, A. Deschamps, A. Simar, *Acta Mater.* **59** (2011) 3621–3635

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