Age Hardening of Aluminium Alloys with T8 Tempering

- Updated in Ver. 16

Introduction

Age Hardening curves of wrought aluminium alloys with consideration of quench sensitivity has been achieved in JMatPro® Ver. 11, combining both strength and precipitation kinetics models. This framework can be found in our earlier report [1]. However, this has only been available for T4/T5/T6 temper designations, where the natural/artificial ageing is applied directly after the initial solution treatment or hot working process. The effect of prior cold work on the strength evolution during subsequent artificial ageing, as exemplified by the T8 temper designation, has now been achieved in JMatPro® Ver. 13. This is by incorporating dislocation density as a new state variable, modelling its evolution during cold work and subsequent ageing, taking into account strain-hardening, dislocation-enhanced precipitation and recovery of the dislocation structure. This report describes this improvement and its validation against experimental data.

Modelling the dislocation density dependent strength evolution

Compared with a T6 treatment, the cold work (e.g. stretching, rolling) in a T8 treatment prior to the artificial ageing mainly has three effects: (1) improving strength prior to ageing via strain hardening; (2) the additional dislocations introduced by cold work could accelerate the precipitation process during subsequent ageing; (3) the enhanced dislocation network can recover (dislocation density decreases) during subsequent ageing, as driven by the climb and annihilation of dislocations at elevated temperatures. This can lead to softening of the material long before the over-ageing stage when precipitate coarsening occurs. All these three aspects would depend on dislocation density, which is incorporated as a new state variable. The evolution of strength as well as the achievable peak strength of a T8 treatment would be the net effect of all the three aspects.

Firstly, after solution treatment or hot forming, the initial yield strength (σ_{y0}) can be expressed as

$$\sigma_{y0} = \sigma_0 + k/\sqrt{d} = \sigma_0 + \alpha MGb\sqrt{\rho_0} \tag{1}$$

where σ_0 is the solid solution strength, $kd^{-0.5}$ is the grain size effect as depicted by the Hall-Petch relationship. The effect of grain size has been understood to be equivalent to the effect of an initial dislocation density ρ_0 after solution treatment or hot forming, as reflected in Eq. (1). G is shear modulus; b is the Burgers vector and M is Taylor factor. α is an obstacle strength parameter [2]. For the strain hardening by cold work, the model takes the original Hollomon strain hardening model $\sigma = K\varepsilon^n$, where the hardening exponent n and strength coefficient K can be calculated from σ_{y0}, and the strain term ε here refers to the percentage of cold work. If $\Delta\rho_1$ is the additional dislocation density generated by cold work, the yield strength after cold work (σ_{y1}) is expressed as

$$\sigma_{y1} = K\varepsilon^n = \sigma_0 + \alpha MGb\sqrt{\rho_0 + \Delta\rho_1} \tag{2}$$

Secondly, precipitation occurs during subsequent artificial ageing. The kinetics of precipitation has been captured by the Johnson-Mehl-Avrami (JMA) theory for the growth and the Lifshitz-Slyozov-Wagner (LSW) theory or Ostwald ripening for the coarsening. Both processes require diffusion of different dominant elements for different phases. Dislocations can provide additional diffusion paths to the precipitation process, through the so-called pipe diffusion, which has been incorporated into the formulation of the effective diffusivity [3]. This treatment allows the effect of deformation to be considered in the precipitation and coarsening processes.

Thirdly, for the recovery of the dislocation network during subsequent ageing, the model proposed by Lagneborg [4] is used, in which the recovery is controlled by a dislocation climb process, causing a reduction of dislocation density $\Delta\rho_2$. The final strength after ageing can then be calculated as

$$\sigma_{y2} = \sigma_0 + \sigma_{PH} + \alpha MGb\sqrt{\rho_0 + \Delta\rho_1 - \Delta\rho_2} \tag{3}$$

where σ_{PH} is the contribution from precipitation strengthening that can be described by Orowan looping or order strengthening mechanisms depending on the precipitate type [5-7].

Evaluating the strength calculation after T8 treatment

In this section, comparisons between the calculated and measured room temperature peak yield strengths of various aluminium alloys following T8 treatment are presented to validate the proposed approach (see Fig. 1). Some of the original data, reported as Vickers hardness values, were converted to yield strength using the internal converter tool in JMatPro®. The heat treatment conditions among the collected datasets vary in the solution-treatment temperature, cooling rate, amount of cold work, and the artificial ageing temperature and duration. Overall, Fig. 1 shows good agreement between the calculated and measured values, with most alloys falling within a ±15% bound.

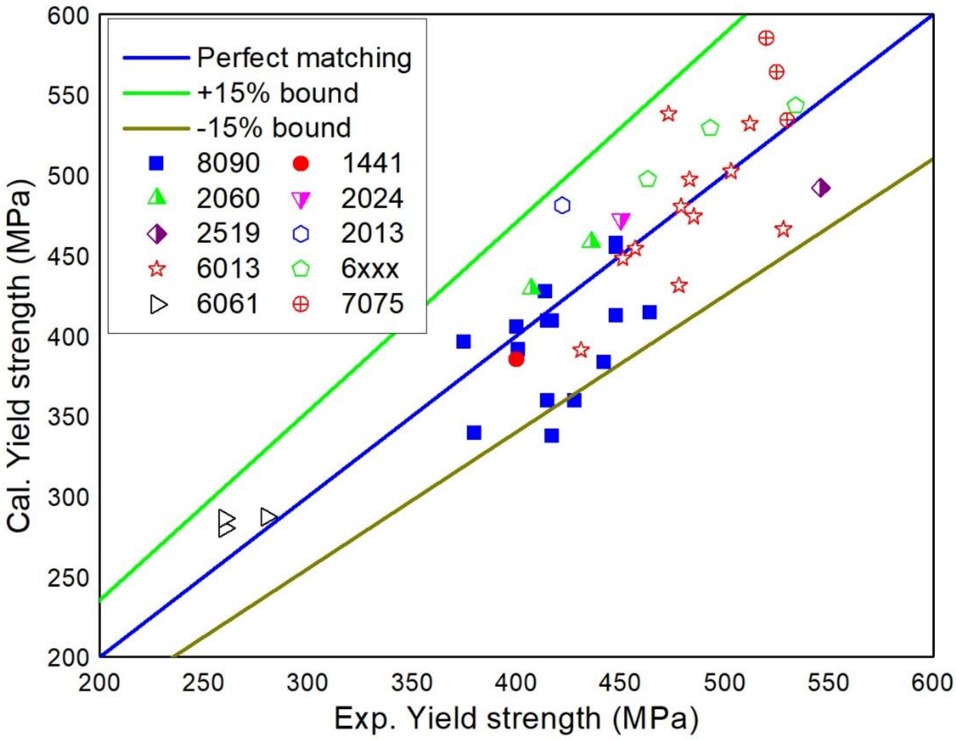
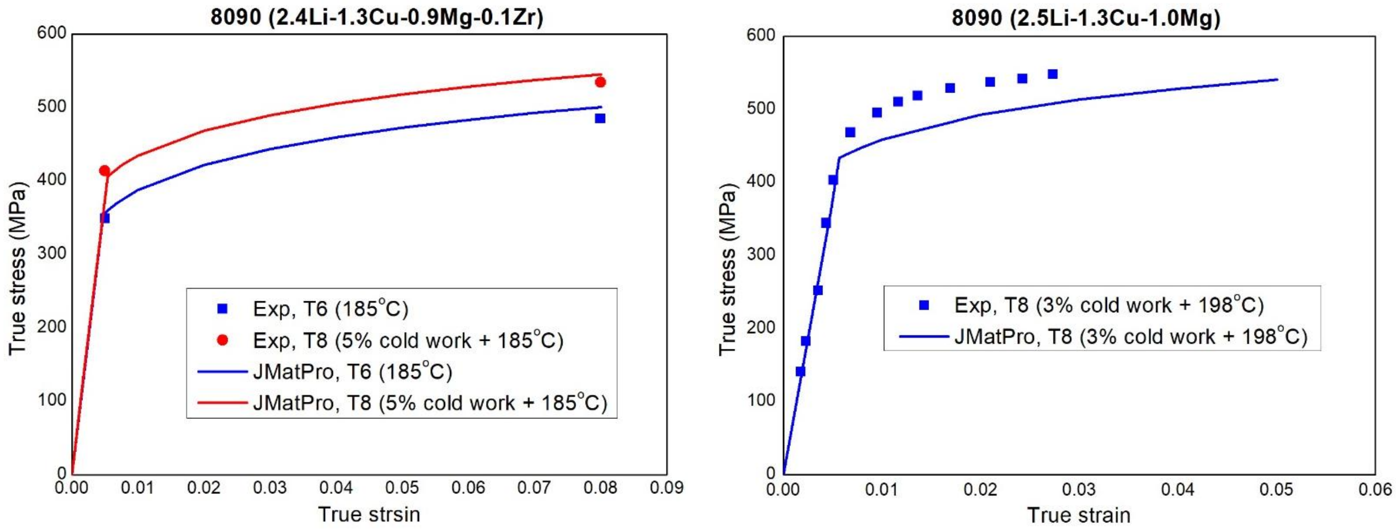


Fig. 1 Comparison between the calculated and measured room temperature yield strengths of different aluminium alloy groups subjected to T8 treatment. Alloy groups and their typical precipitate phases are shown as follows: 2060, 2024 and 2519 (*S'* and *θ'*) [8-12]; 2013, 6013, 6xxx and 6061 (*B'*, *β'*, *β''* and *Q'*) [13-17]; 8090 and 1441 (Al_3Li) [8,18-20] and 7075 (*η'*) [11,21,22].

Fig. 2 presents the flow stress curves of two 8090 alloys. The good agreement with data further quantitatively demonstrates that a small percentage of cold work in the T8 treatment can significantly improve the strength at the room temperature, as compared with the T6 treatment counterpart with the same ageing temperature and time.



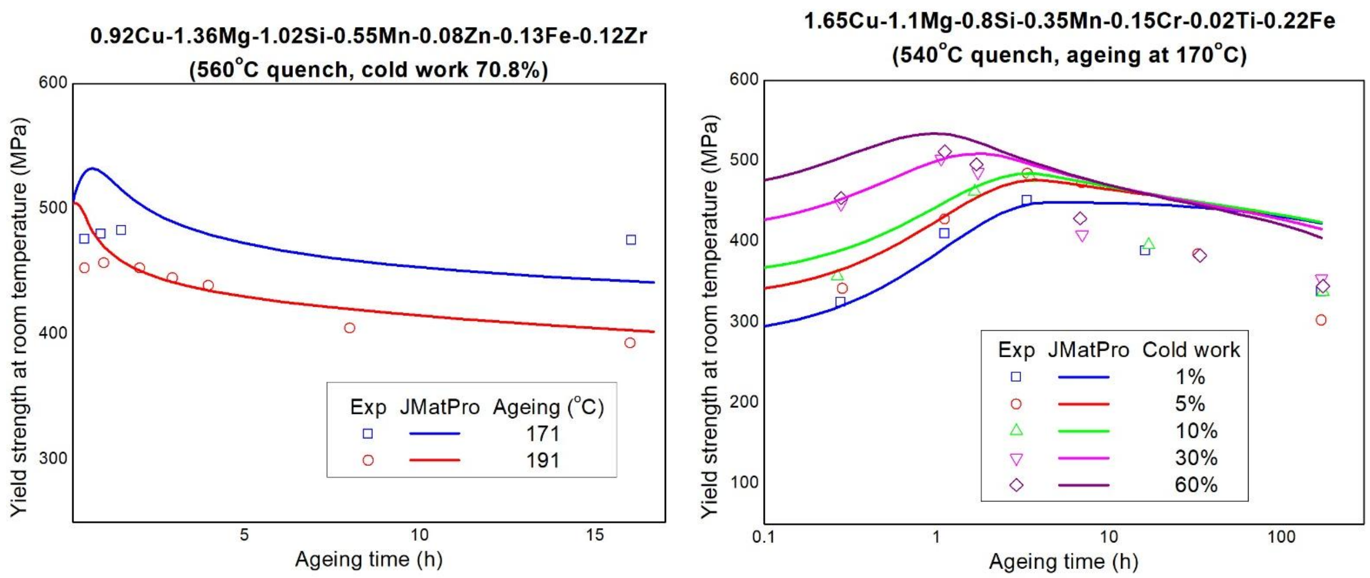
(a) (b)

Fig. 2 The flow stress curves of two different 8090 alloys, subjected to different cold work percentages and heat treatments. (a) 2.4Li-1.3Cu-0.9Mg-0.1Zr wt% [8]; (b) 2.5Li-1.3Cu-1.0Mg wt% [18].

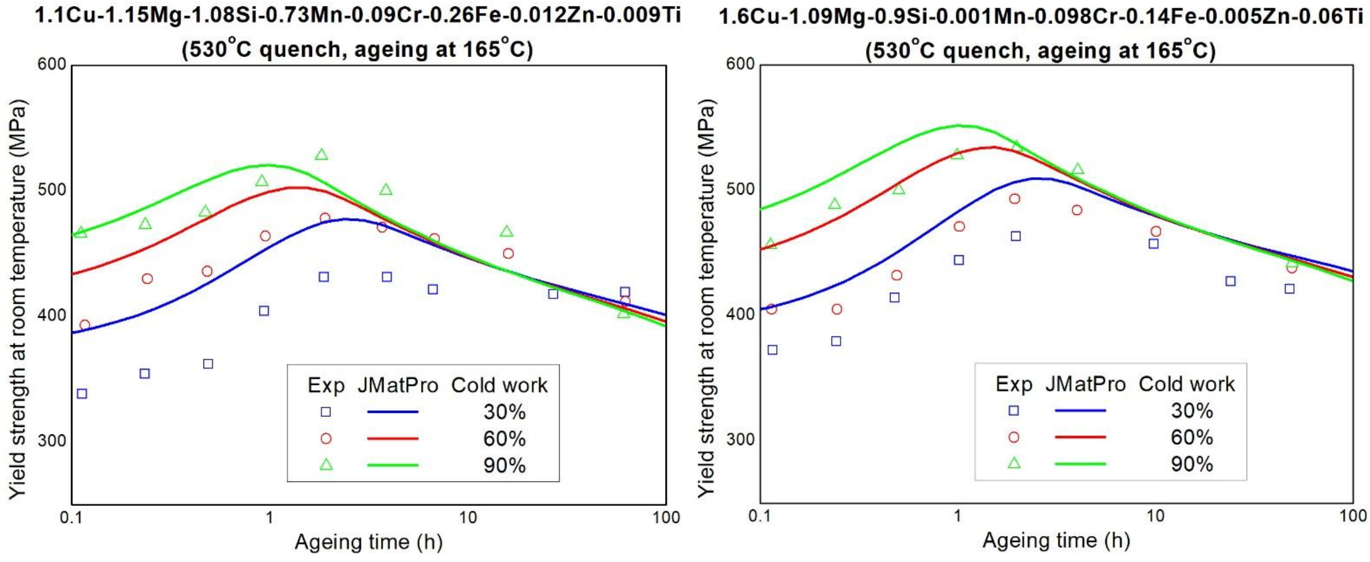
Figure 3 presents detailed strength evolution curves for several 6xxx alloys. The influence of ageing temperature is shown in Fig. 3(a), which demonstrates that the peak strength tends to decrease as the ageing temperature increases, and higher ageing temperatures also accelerate recovery during ageing. Figures 3(b)-(d) illustrate the effect of varying cold-work levels. A higher cold work percentage increases the initial dislocation density prior to ageing, resulting in a faster rise to a higher peak strength.

The data in Fig. 3(c) and (d) are collected from the same study [16], which investigated the effect of alloy composition on strength evolution. Under identical heat treatment conditions, the higher peak strength in Fig. 3(d) compared with Fig. 3(c) is well captured by JMatPro®. This difference can be attributed to a higher fraction of metastable phases in the alloy shown in Fig. 3(d) at the ageing temperature (165°C).

It is also noteworthy that all curves in Fig. 3(b)–(d) converge at later ageing stages. This occurs because the additional dislocation strengthening disappears due to recovery, leaving precipitate coarsening as the dominant strengthening mechanism.



(a) (b)



(c) (d)

Fig. 3 Comparison of the strength evolution of 6xxx alloys: (a) 0.92Cu-1.36Mg-1.02Si-0.55Mn-0.08Zn-0.13Fe-0.12Zr wt%, subjected to different ageing treatments [14]. (b) 1.1Cu-1.15Mg-1.08Si-0.73Mn-0.09Cr-0.26Fe-0.012Zn-0.009Ti wt% [15]. (c) 1.65Cu-1.1Mg-0.8Si-0.35Mn-0.15Cr-0.02Ti-0.22Fe wt% [16]. (d) 1.6Cu-1.09Mg-0.9Si-0.001Mn-0.098Cr-0.14Fe-0.005Zn-0.06Ti wt% [16]. Alloys in (b)-(d) are subjected to different cold work percentages.

Summary

The influence of prior cold work on strength evolution during subsequent artificial ageing of aluminium alloys, as exemplified by the T8 temper designation, has been investigated and incorporated as an enhancement to the existing Age Hardening module in JMatPro®. Dislocation density has been introduced as an additional state variable to quantify the effect of cold work, accounting for three key aspects: additional strain hardening, accelerated precipitation and dislocation network recovery.

The approach has been validated by good agreement with extensive literature data on room temperature yield strength after T8 treatment across various aluminium alloy series, each dominated by different types of precipitate phases. It captures the fact that cold work can increase peak strength and accelerate its achievement. It also highlights the role of dislocation network recovery in the reduction of strength after the achievement of peak strength, which in many cases outweighs the effect of precipitate coarsening. Ultimately, this improvement could assist the design and assessment of processing methods in the aluminium industry.

References

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