Age Hardening of Aluminium Alloys with T8 Tempering

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

$$\sigma_{y0} = \sigma_0 + k/\sqrt{d} = \sigma_0 + \alpha M G b \sqrt{\rho_0} \qquad (1)$$

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], and the value of ρ_0 is typically 10^{10}~$10^{12}/\mathrm{m}^2$. 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 $\Delta\rho$ is the additional dislocation density generated by cold work, the yield strength after cold work (σ_{y1}) is expressed as

$$\sigma_{y1} = K\varepsilon^n = \sigma_0 + \alpha M G b \sqrt{\rho_0 + \Delta\rho} \qquad (2)$$

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.

$$\frac{d\rho}{dt} = -2\theta Q \Gamma \rho^2 = -\frac{\theta D_s G b^3}{kT}\rho^2 \qquad (3)$$

where $Q = D_s b/kT$ is the mobility of the climbing dislocation, k is Boltzmann's constant, D_s is the lattice diffusivity of Al in the matrix. $\Gamma = 1/2Gb^2$ is the dislocation line tension. θ is introduced here as a fitting constant. Note since there is no strain hardening involved in this recovery process, an explicit solution of the final value of ρ can be obtained as below, with $\rho_0 + \Delta\rho$ after cold work being the starting point.

$$\rho = \frac{1}{At + 1/(\rho_0 + \Delta\rho)} \tag{4}$$

where $A = \theta D_s G b^3/kT$ is considered as a recovery constant. The final strength after ageing can then be calculated as

$$\sigma_{y2} = \sigma_0 + \sigma_{PH} + \alpha M G b\sqrt{\rho} \tag{5}$$

where σ_{PH} is the contribution from precipitation that can be described and quantified 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 yield strengths of different aluminium alloys after T8 treatment are presented to validate the proposed approach. Note that some original data are in the form of room temperature Vickers hardness. These have been converted to yield strength using the JMatPro® internal converter tool. The heat treatment of the collected data varies in the solution treatment temperature, cooling rate, percentage of cold work, as well as the artificial ageing temperature and time. It should be noted that the cold deformation of some data has reached up to 90%. A general comparison of the T8 peak strength is shown in Fig. 1. Good agreement can be seen in the figure, which reveals a ±10% bound region for the majority of alloys.

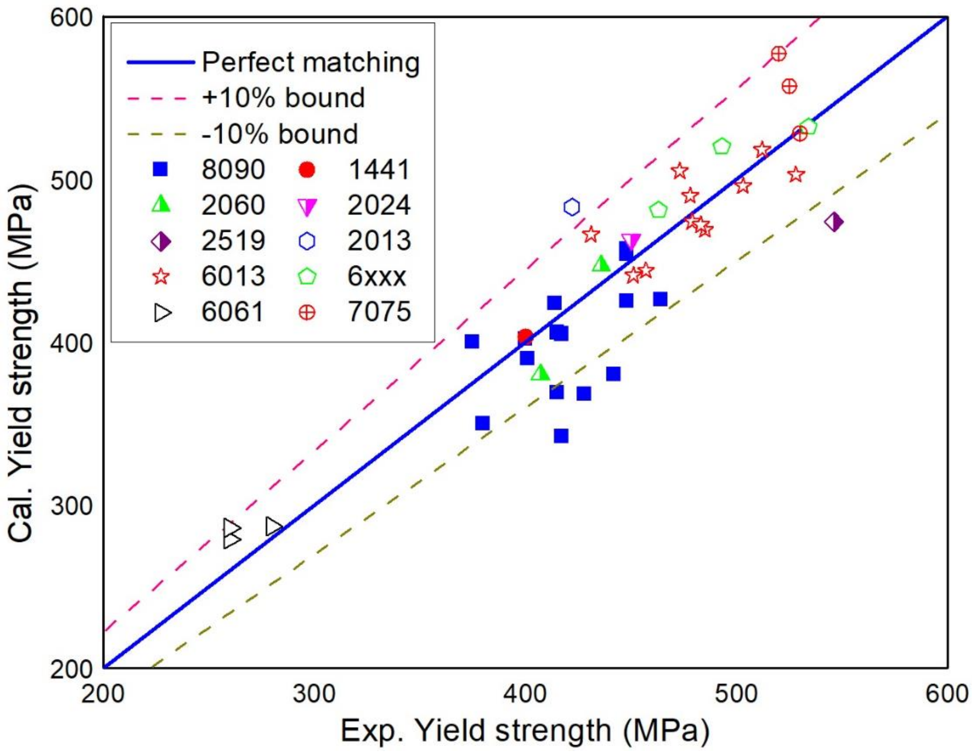
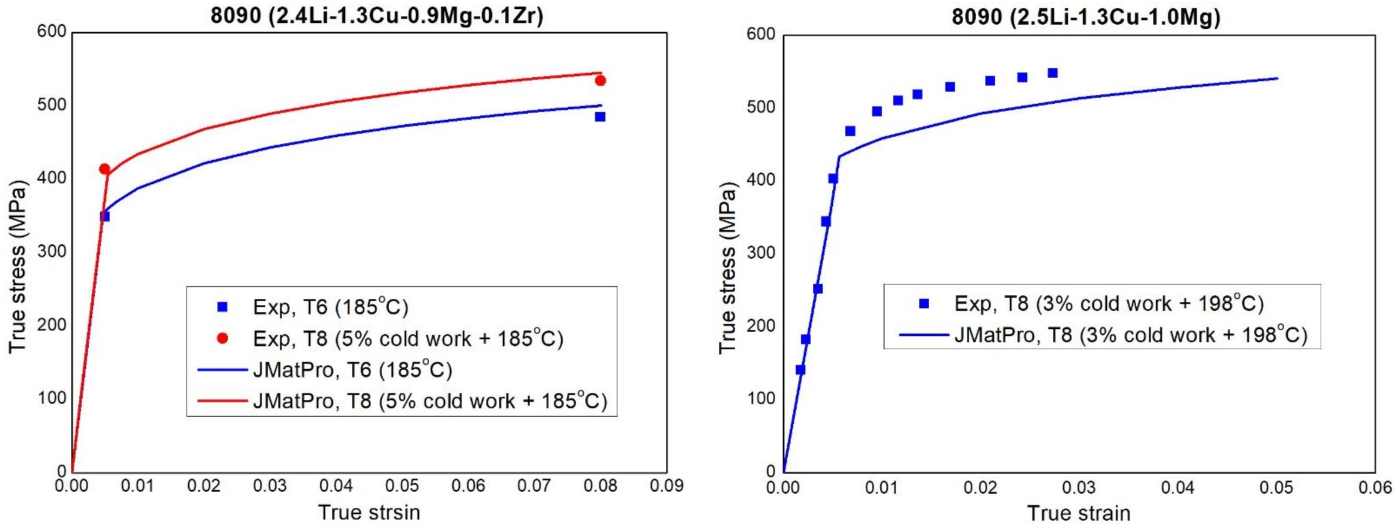


Fig. 1 Comparison between the calculated and measured room temperature yield strength of different groups of aluminium alloys subjected to T8 treatment. 2060, 2024 and 2519 have typical phases of S' and θ' [8-12]; 2013, 6013, 6xxx and 6061 have typical phases of B', β', β" and Q' [13-17]; 8090 and 1441 have typical phases of Al_3Li [8,18-20] and 7075 has a typical phase of η' [11,21,22].

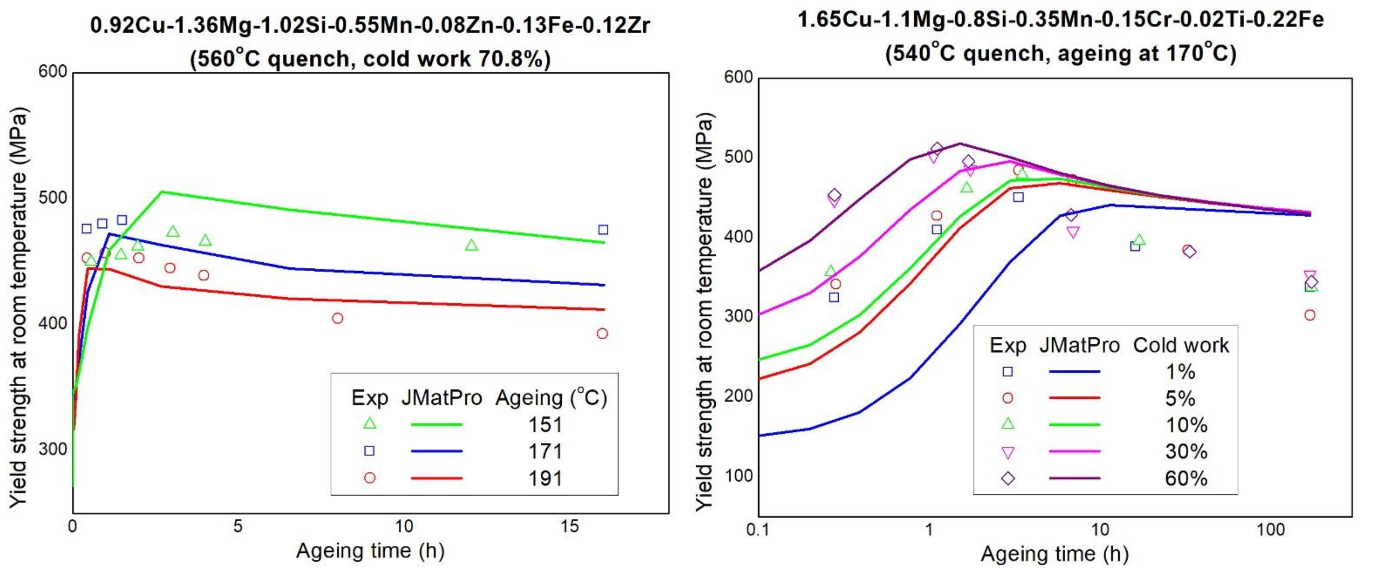
Fig. 2 presents the flow stress curves of two 8090 alloys. The good agreement with data further quantitatively demonstrates that a small percentage of cold work in the T8 treatment can significantly improve the strength at the room temperature, as compared with the T6 treatment counterpart with the same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



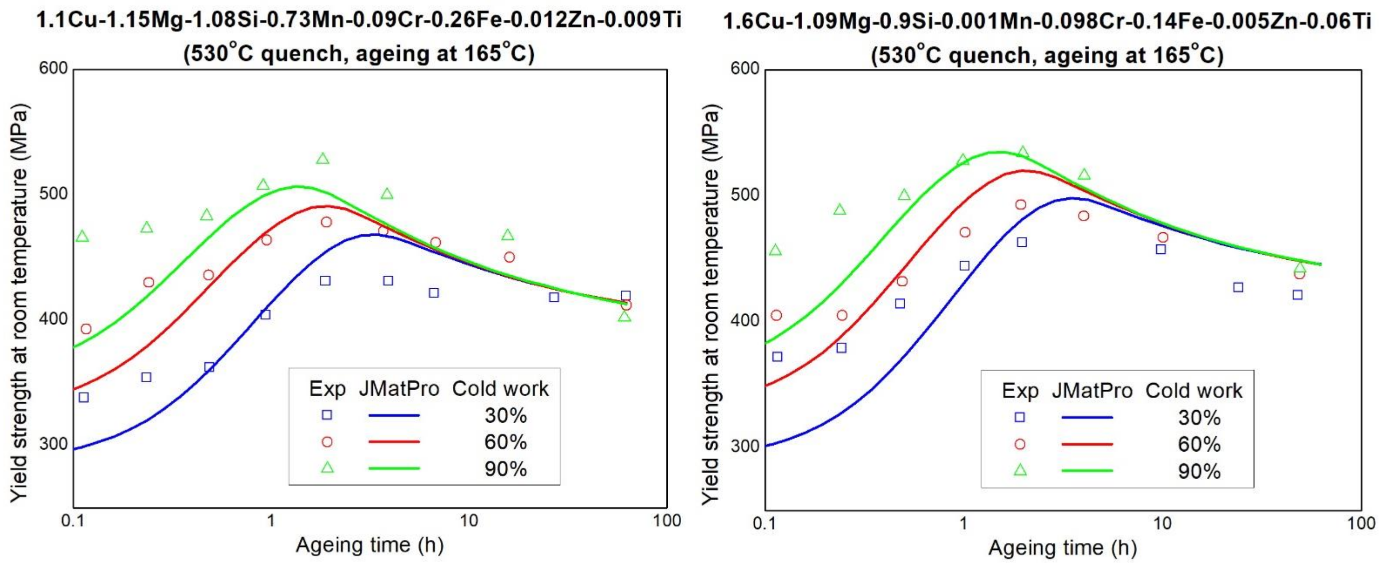
(a) (b)

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].

Some detailed strength evolution curves of 6xxx alloys are presented in Fig. 3. The effect of different ageing temperatures is shown in Fig. 3(a), which demonstrates that the peak strength may decrease with the increase of the ageing temperature and a higher ageing temperature can lead to a faster recovery during ageing (Eqs. 3&4). The effect of different percentages of cold work is reflected in Fig. 3(b)-(d), where a higher percentage of cold work can give rise to a higher initial dislocation density before ageing (Eq. 2) and lead to faster achievement of a higher peak strength. Data in Fig. 3(c) and (d) are collected from the same work [16], which aimed to reveal the effect of composition on the strength evolution. With the same heat treatment condition, the higher peak strength in Fig. 3(c) compared with its counterpart in Fig. 3(d) is well captured by JMatPro®, which can be attributed to a higher fraction of metastable Q' phase for the alloy in Fig. 3(c) at the given ageing temperature (165°C). Note that all the curves in Fig. 3(b)-(d) appear to coincide at the later stage of ageing, which is when the effect of additional dislocations disappears by recovery, leaving the subsequent dominant strengthening mechanism being the coarsening of precipitates.



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

The approach has been validated by a good agreement with extensive data in literature of the measurement of room temperature yield strength after T8 treatment across different series of aluminium alloys that are dominated by different types of precipitate phases. It enables to capture the fact that cold work can help enhance the 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 contribution by any coarsening of precipitates. Ultimately, this improvement could assist the design and assessment of processing methods in the aluminium industry.

References

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