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PROCESS MODEL BASED OPTIMISATION OF COOLING SCHEDULES FOR AA6082 EXTRUSIONS

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Abstract

The aluminium alloy AA6082, which is frequently used for extruded products, is known to be quench sensitive. This implies that too slow a cooling leads to a reduction of several important properties such as strength, formability, fatigue strength, toughness, corrosion resistance and surface appearance. Rapid cooling, however, is also detrimental as this causes distortion of the extruded sections, which can be costly or impossible to rectify.

The present paper shows how process modelling techniques can be applied to optimise the cooling schedule for AA6082 extrusions. The model predicts the strength of the alloy after continuous cooling and artificial ageing, and is based on well-established principles of thermodynamics, kinetic theory and simple dislocation mechanics. A brief summary of the model is given here, and the results of a range of predictions are presented. These show that the peak hardness is a function both of the cooling rate, the intrinsic dispersoid density of the material, and the Mg and Si composition, in agreement with general experience.

Introduction

The age hardening aluminium alloys have been widely studied, to the extent that most of the underlying physical processes are well-established [1-4]. In recent years, successful attempts have been made to assemble complete process models for the behaviour of Al-Mg-Si alloys during heat treatment and welding [5-7]. The present paper deals with modelling of the quench sensitivity of alloy AA6082, which is frequently hot formed by extrusion or forging. Solution treatment is achieved during the forming process, but the subsequent quench is critical. This is because β'-Mg_2Si precipitation can occur during cooling at Mn-based dispersoids, leading to reduced strength after ageing to the T5 temper [1-4].

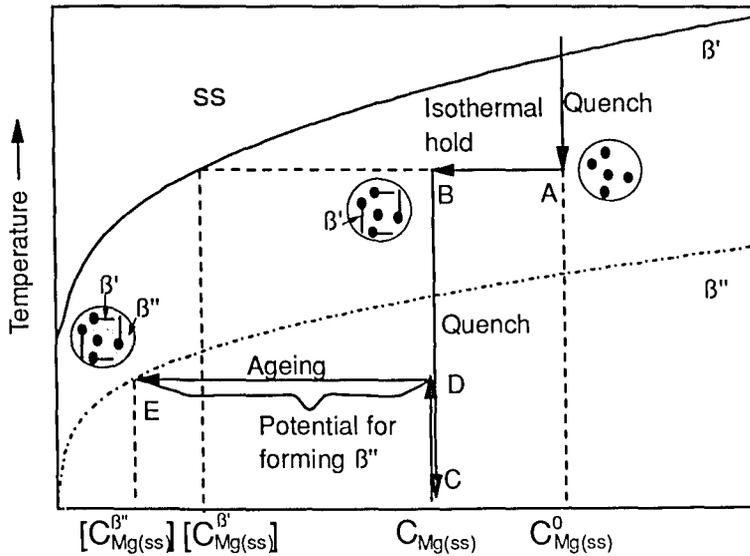**Fig. 1.** Schematic representation of the microstructure evolution during cooling and subsequent artificial ageing referred to the phase diagram.

Basis of the model

Modelling of quench sensitivity essentially requires a description of the evolution of solute in the matrix during cooling and subsequent ageing. In common with most phase transformation models, isothermal behaviour is first described and then continuous cooling is dealt with by appropriate integration. Figure 1 shows schematically the evolution of matrix solute content during an interrupted quench and subsequent ageing, with respect to the phase diagram. Starting with a fully supersaturated solid solution (A), precipitation of ß' depletes the matrix of solute (A-B), leaving a reduced supersaturation on quenching to room temperature (C). The potential for ageing now depends on the relative levels of retained solute, and the equilibrium solute for ß"-Mg$_2$Si precipitation (D and E). A mathematical description of this sequence therefore requires; (i) a thermodynamic model for the solvus boundaries, (ii) an isothermal kinetic model for ß' precipitation (and its adaptation for continuous cooling), and (iii) a model for the precipitation strength increment during artificial ageing. These are summarised briefly here. Full details will be published elsewhere [8,9].

Thermodynamic model

In the present modelling exercise, the ß, ß' and ß" precipitates are all assumed to be stoichiometric Mg$_2$Si. The stable and metastable phase boundaries may therefore be described using the solubility product for Mg$_2$Si:

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$$T_{eq} = \frac{5010 - \Omega}{5.85 - \log[C_{Mg(ss)}]_r^2 [C_{Si(ss)}]_r} \tag{1}$$

where Ω is the contribution of the interface curvature to the reaction enthalpy (given by the Gibbs-Thomson equation [10]). Hence $\Omega = 0$ describes the stable ß solvus boundary.

Isothermal kinetic model

In general, the volume fraction of ß'-Mg_2Si that precipitates at dispersoids by quenching to an intermediate hold temperature can be calculated from the Avrami-equation [2,7]:

$$f_{\beta'} = f_{\beta'}^{eq} \left(1 - \exp\left(-\left(\frac{t}{\tau}\right)^n \right) \right) \tag{2}$$

where $f_{\beta'}^{eq}$ is the equilibrium volume fraction of ß' at the hold temperature, τ is a temperature-dependent time constant, and n is a time exponent. The volume fraction of ß' is directly related to the fraction of available solute which remains after quenching to room temperature at the end of the isothermal hold. We can therefore define the dimensionless fraction transformed as:

$$X_{Mg}' = \frac{f_{\beta'}}{f_{\beta'}^{eq}} = \frac{C_{Mg(ss)}^0 - C_{Mg(ss)}}{C_{Mg(ss)}^0 - [C_{Mg(ss)}^{\beta'}]} \tag{3}$$

where $[C_{Mg(ss)}^{\beta'}]$ is the equilibrium concentration of Mg in solid solution at the hold temperature. In equation (3) the solute concentration is referred to as $C_{Mg(ss)}$, but due to the stoichiometry of the Mg_2Si reaction, this quantity essentially includes the appropriate proportion of Si. Experimental measurements of X'_{Mg} are made using electrical resistivity (taking due account of excess Si and other alloying elements in solution). Full details are given in Ref. 9.

Manipulation of equation (2) yields an alternative convenient form for X'_{Mg}:

$$X_{Mg}' = 1 - (1 - X_c)^{\left(\frac{t}{t^*}\right)^*} \tag{4}$$

where t^* denotes the time for a particular fraction X_c to transform at a given temperature. The form of $t^*(T)$, which provides a full mathematical description of the C-curve in temperature-time space, is [8]:

$$t^* = t_r^* \left(\frac{N_v^*}{N_v}\right) \exp\left[\frac{1}{R}\left(\frac{A_0(T_{eq})^2}{T(T_{eq} - T)^2} - \frac{A_0^*(T_{eq}^*)^2}{T_r(T_{eq}^* - T_r)^2} \right) + \frac{Q_d}{R}\left(\frac{1}{T} - \frac{1}{T_r} \right) \right] \tag{5}$$

The solvus boundary (i.e., T_{eq} or T^*_{eq}) for ß' is calculated from equation (1) while calibration of the other constants in this equation (all of which have a physical basis) is achieved using resistivity data after a range of quench-hold-quench treatments. Figure 2 shows a set of data for
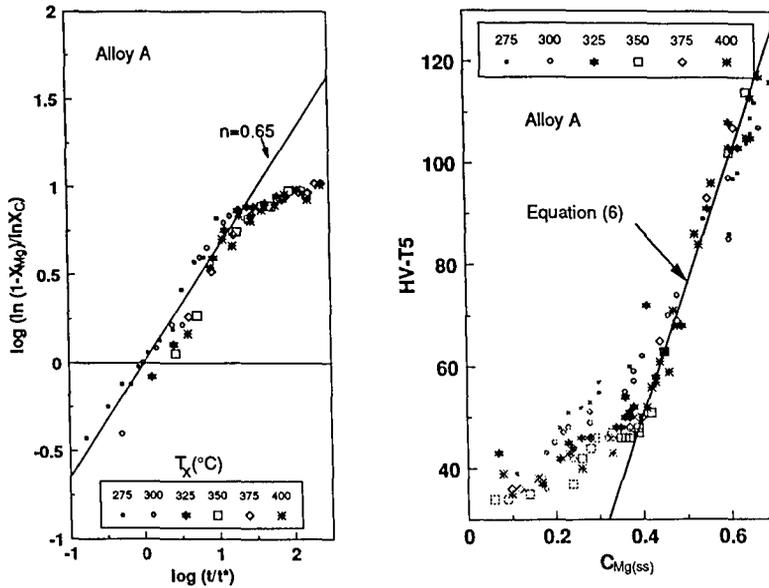
Fig. 2. Reaction kinetics of AA6082 based on Fig. 3. Experimental verification of eqn. (6).
the Avrami equation.

fraction transformed against normalised hold time at various interrupt temperatures for one AA6082 alloy. The correlation with the theory is good. In all, four different AA6082 alloys have been examined, varying the dispersoid density (by changing homogenisation temperature) and the Si content. Similar agreement is obtained in all cases, by adjustment of a few alloy-dependent parameters.

Precipitation strength model

Assuming that the strengthening effect of ß"-Mg₂Si is associated with order hardening, the precipitation strengthening increment ΔS_p is proportional to the volume fraction of ß" particles [11]. This, in turn, is proportional to the supersaturation following the quench, and the equilibrium solute level for ß" (Fig. 1). A dimensionless strength increment may thus be defined:

$$\alpha = \frac{S - S_{min}}{S_{max} - S_{min}} = \frac{\Delta S_p}{\Delta S_p(\text{max})} = \frac{C_{Mg(ss)} - [C^{\beta''}_{Mg(ss)}]}{C^0_{Mg(ss)} - [C^{\beta''}_{Mg(ss)}]} = X''_{Mg} \qquad (6)$$

where X''_{Mg} is calculated on the basis of resistivity after the quench [9], and S_{max} and S_{min} are upper and lower T5 strengths measured after subsequent artificial ageing. S_{max} corresponds to a rapid quench (with the full potential for T6 ageing), and S_{min} to a slow quench (after which

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no supersaturation with respect to β" exists). Figure 3 shows the T5 hardness against measured X''_{M_g} for the same quench-hold-quench experiments as in Figure 2 to confirm the general validity of equation (6).

Application to continuous cooling and real extrusions

The isothermal model is a single state variable model describing the fraction of solute lost to ß' precipitates. Isothermal models such as this may, in general, be applied to non-isothermal conditions if the additivity requirements for an isokinetic reaction are satisfied [12]. If these requirements are not met, then stepwise integration of the Avrami equation (in differential form) is required. The implications of these matters have been addressed in detail elsewhere [8,9] and
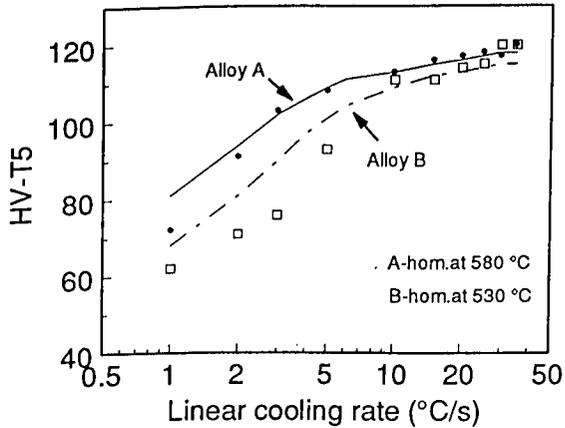

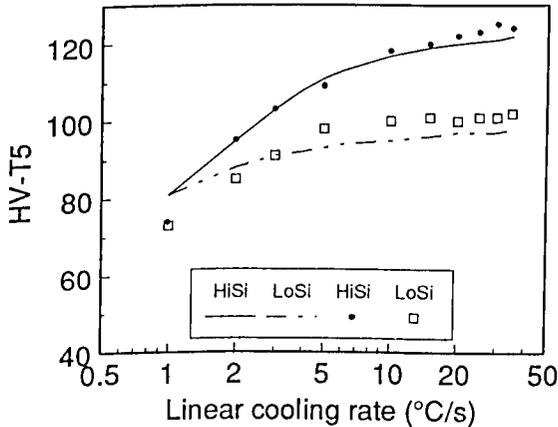**Fig. 4a.** Application of model to continuous cooling; Effect of dispersoid density.

Fig. 4b. Application of model to continuous cooling; Effect of Si content.

will not be elaborated here. It is sufficient to note that for a given cooling history, the model can be integrated appropriately, leading to predictions of as-quenched supersaturation and hence T5 strength. Figure 4 shows the model predictions for cooling at a constant rate, compared with experimental data (mostly obtained from specimens treated in a weld simulator). Figure 4a shows the data for AA6082 containing 2 different dispersoid densities, while Fig. 4b shows the data for AA6082 alloys with two extreme Si contents. The agreement with the models is excellent. It should be noted that for each alloy the material constants have been tuned to quench-hold-quench data, and then applied directly to continuous cooling with no further adjustment.

Application to industrial extrusions

Thermal histories can be obtained from industrial extrusions at various levels of precision. Complex finite element (FE) computations are made for a wide range of purposes (die design, prediction of distortion, microstructure prediction) [13]. These yield detailed cooling histories for all parts of the extrusion, limited primarily by the complexity of modelling surface heat transfer. Alternatively, approximate analytical solutions may be derived (and checked against the FE predictions) which capture the cooling rate. These solutions, coupled with the model for the microstructural evolution, have sufficient accuracy to estimate ·the T5 strength when extrusions of a given shape are quenched by different means (e.g. water, air or spray). Tjøtta [14] describes the derivation of an approximate equation for the temperature history of an extrusion leaving the press:

$$T(t) \approx (T_{init} - T_c) \exp\left(-\frac{t}{t_{ref}}\right) + T_c \qquad (7)$$

The important parameter is t_{ref}, which takes into account the profile geometry (based on the ratio
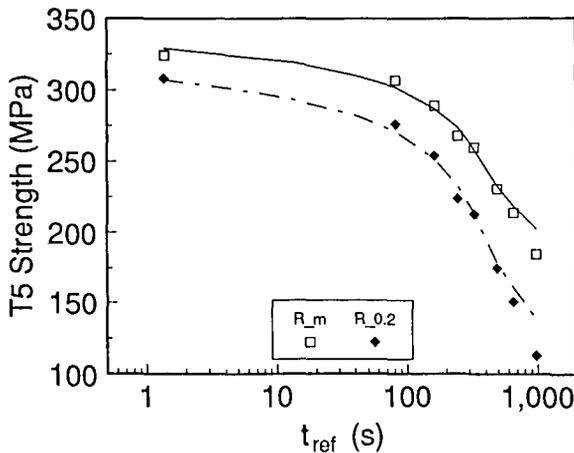

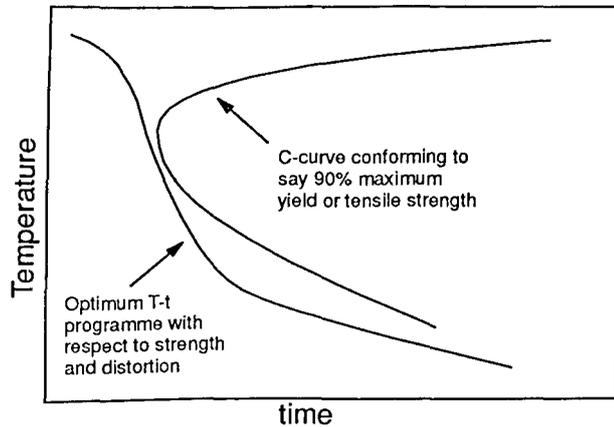Fig. 5. Application of the model to different extruded profiles.

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Fig. 6. Optimalisation of cooling schedule for AA6082 extrusions (schematic).

of surface area to cross-sectional area), the physical properties of the material and the heat transfer coefficient. We have used this approximate cooling model to predict the T5 strength of a range of representative sections from production. Figure 5 shows the predictions of the model for different sections and cooling conditions, with the corresponding approximate values of t_{ref}. The problem here is not the yield or tensile strength itself, since the strength requirement can easily be met by the use of for instance forced water-cooling. However, because rapid cooling may cause serious distortion of the extrusions, optimalisation of the cooling schedule is essential in order to achieve the desired quality of the products. In practice, this requires careful control of the temperature-time programme, as illustrated in Fig. 6. The advantage of the present model is that it may easily be implemented in a spreadsheet and used online for evaluating the effect of different cooling systems on the strength of a given extruded profile.

Conclusion

It is demonstrated that process modelling techniques can be applied to describe the quench sensitivity of Al-Mg-Si extrusions. The combination of a thermodynamic model and an isothermal kinetic model gives a good prediction of the loss of solute to precipitates at dispersoids, and the consequent reduced T5 strength. The model may be applied to continuous cooling, in particular to cooling of real industrial extrusions. The combination of the microstructure model with an approximate model for heat flow creates a powerful tool for designing cooling systems for extrusion operations.

Acknowledgements

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