The Homogenisation, Extrusion and Ageing of a Commercial 2xxx Series (Al-Cu) Aluminium Alloy

M.J. Lawday, M.R. Clinch, S.J. Harris, B. Noble

School of Mechanical, Materials, Manufacturing Engineering and Management,
University of Nottingham, University Park, Nottingham, NG7 2RD, UK

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Abstract

A systematic study has been conducted to produce a commercially feasible practice for the homogenisation of a 2xxx series (Al-Cu) aluminium alloy. Hardness was shown to be a very satisfactory index of workability, with a maximum value of 65HV10 specified to enable the production of defect free components by a cold extrusion type process. Subsequent heat treatment of the formed parts to the T6 condition allowed the development of a high strength alloy. The study has considered the following factors which may influence workability: homogenisation temperature, homogenisation time, cooling rate to ambient and natural ageing during storage. The main inter-metallic was identified as θ-phase ($CuAl_2$). The volume fraction remaining out of solid solution was measured for extrusion by optical image analysis, and confirmed by differential scanning calorimetry (DSC).

1. Introduction

Al-Cu and Al-Cu-Mg (2xxx series) aluminium alloys have been used for structural applications in the aerospace industry [1]. In recent years, the number of applications for commercial 2xxx series alloys has increased to include highly competitive sectors such as automotive and packaging, where their high specific mechanical properties are also important. There has been an increased demand for both alloy and process optimisation, in order to reduce the lead time to market for new products in these sectors. Whilst this has lead to a surge of numerical and physical based modelling [2], the work required to generate and validate a model is still, at present, far greater than that required for previous approaches to optimisation.

The present study considers the systematic development of a commercially feasible practice for the homogenisation of 2xxx series aluminium alloy (Al-5.6%Cu-0.3%Mg) DC-cast billets used in the production of light-weight, high-strength structural components. Whilst workability is a complicated engineering phenomenon [3], this study attempts to find an indication of satisfactory workability which will permit the production of components by cold extrusion without defects. Subsequent heat treatment of the formed parts to the T6 condition then allows the development of a high strength alloy. The study considered the following factors which may influence workability: homogenisation temperature, homogenisation time, cooling rate to ambient and natural ageing during storage.

2. Experimental Procedure

DC-cast 2xxx series aluminium alloy billets (150Ø, 3m in length), with a composition described in table 1, were used in this investigation. The composition lies in the $\alpha + \theta + Q$ ($\alpha + CuAl_2 + Cu_2Mg_8Si_6Al_5$) region of the 460°C isothermal section of the Al-Cu-Mg-Si quaternary phase diagram [4] (Figure 1). Metallographic examination in the as-cast condition was undertaken using optical (Nikon ME600) and scanning electron microscopy (Jeol 6400 SEM, 20kV beam). Differential scanning calorimetry (Perkin Elmer 7 DSC) at a scanning rate of 20°C per minute was used to study solid-state reactions, whilst optical image analysis (using Aequitas® IA Lite software) was used to measure volume fractions of eutectic.

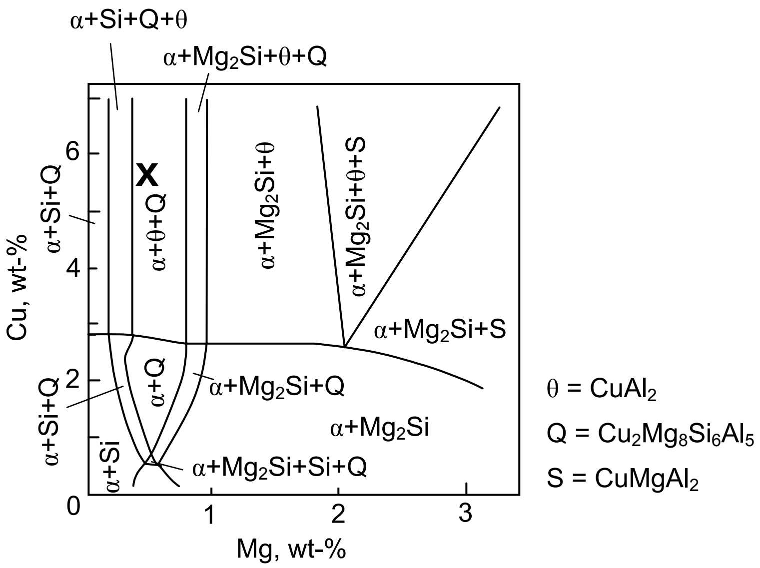

Figure 1: Isothermal section at 460°C of the Al-Cu-Mg-Si quaternary phase diagram for 0.6%Si. The chemical composition of the AA2001 aluminium alloy studied is indicated [4].

Table 1: Chemical composition of the 2xxx series aluminium alloy studied (wt.%)

Cu	Si	Mg	Mn	Fe	Cr
5.6	0.11	0.2-0.45	0.25	0.08	<0.005

The process steps followed throughout this study are shown in figure 2. Billet sample slices (edge to centre) were homogenised in a fan assisted laboratory oven with an 8-step programmer. Saturation temperatures of 400, 500, 525, 535 and 545°C were obtained with a ramp time of 2h, to ensure dissolution of the low temperature Q-phase ($Cu_2Mg_8Si_6Al_5$) [5]. Samples were subsequently held for 0-24h at temperature, prior to a water quench. On removal from the quench bath, DSC and optical image analysis was used to measure volume fractions of eutectic.

Two homogenisation practices, A-DC (Direct Cool) and B-SFC (Step Furnace Cool), based on the saturation temperatures of 525 and 400°C respectively, were used initially to assess the influence of cooling rate on workability, and subsequent susceptibility to natural ageing prior to extrusion. Three further practises were later developed for investigation: A-SFC, designed using C-curve data for AA2017 [6], to investigate the effect of a stepped furnace cool from 525°C, and A-FC and B-DC, to investigate a furnace cool and direct cool from 525 and 400°C respectively. Both the step furnace cooled practises are shown in figure 3. Vickers hardness tests (10kg) were made daily on the samples over a two week period. Torsion tests were also undertaken on a TecQuipment manual torsion test machine at one week intervals and compared to samples in the as-cast condition. The applied torque and twist angles were converted into shear stress and shear strain, which were subsequently converted into equivalent stress-strain (σ_{eq}-ε_{eq}) curves [7].

70 billets were homogenised using either A-SFC or B-SFC practices prior to deformation via a cold extrusion type process. The extruded components were subsequently aged to a T6 temper. Ultimate tensile strength, proof strength and elongation values were obtained via tensile testing.

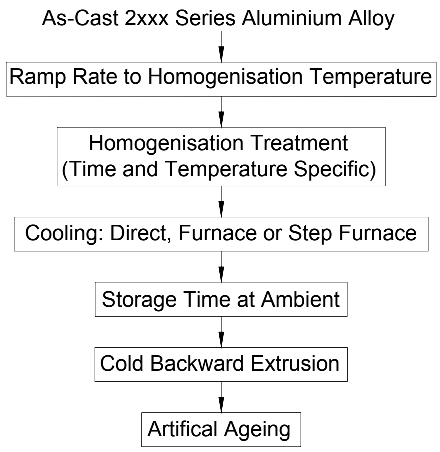

Figure 2: Process outline.

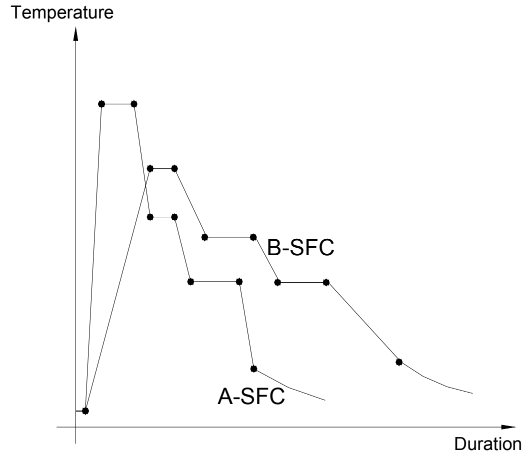

Figure 3: Schematic of the step furnace cool homogenisation practices.

3. Results and Discussion

3.1 As-Cast Material Characterisation

The presence of both Q ($Cu_2Mg_8Si_5Al_5$) and θ ($CuAl_2$) phase was confirmed in the as-cast condition by differential scanning calorimetry (DSC), with endothermic dissolution peaks occurring at 510 and 548°C respectively, which agrees with that of α + $CuAl_2$ eutectic, and α + $CuAl_2$ + $Cu_2Mg_8Si_6Al_5$ as stated by Mondolfo [5]. X-Ray Diffraction (XRD) and Energy Dispersive Spectroscopy (EDS) analysis further confirmed the main segregate to be $CuAl_2$, however the ternary eutectic could not be detected using these techniques due to its small volume fraction (<0.5%). The presence of this eutectic in the as-cast condition gives rise to a material with an average hardness of 90 HV10.

3.2 Homogenisation Temperature and Time

An initial homogenisation practice, with a maximum temperature of 500°C achieved at a rate of 250°C h^{-1} allowed sufficient time for the majority of the soluble intermetallic to dissolve. DSC results showed that after the 2h ramp, the ternary eutectic ($Cu_2Mg_8Si_5Al_5$) had completely disappeared but some $CuAl_2$ phase was still present. The dissolution of the ternary eutectic in the ramp up-period was attributed to its small initial volume fraction, combined with the final homogenisation temperature being close to its melting point, thus providing a sufficiently high driving force for a fast diffusion rate. As a result the ternary eutectic was not considered commercially significant provided a suitable ramp-up period is given during homogenisation.

After a homogenisation treatment of 4h at 400, 500, 525, 535 and 545°C, the enthalpy (ΔH) for the dissolution of θ-phase was observed to decrease from 24.5Jg^{-1} in the as-cast condition, to 17.5, 14.3, 6.2, 2.6 and 1.2Jg^{-1} respectively (Figure 4). Increasing the homogenisation time beyond 4h had no further significant effect on the volume fraction of eutectic that remained undissolved, thus indicating 4h to be the optimum homogenisation dwell time. The DSC results correlated closely with those obtained by optical image analysis, which indicated the volume fraction of eutectic in the as-cast material to be 11.3%, and to subsequently decrease to 5.4, and 2.3% after 4h at 400 and 525°C respectively (Figure 5a-c). This difference in volume fraction of eutectic at the respective temperatures agrees with results reported in the literature [3,8-9] and corresponds closely to the Al-Cu binary phase diagram [5]. It should be noted that samples homogenised at 545°C showed significant porosity (Figure 5d). Since this was not present in any of the lower temperature homogenisation treatments, the high level of porosity was attributed to the incipient melting of some eutectic phase. This indicated that the composition of the alloy (5.6wt%Cu) was too close to the composition of the $CuAl_2$ eutectic line (5.7wt%Cu [5]) for complete dissolution to occur. A maximum homogenisation temperature of 525°C was therefore set to avoid the risk of incipient melting.

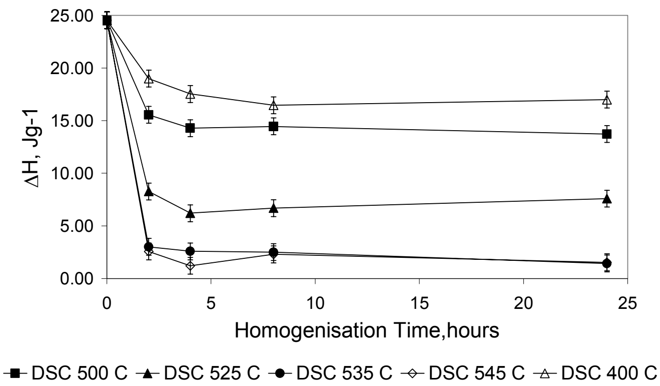

Figure 4: ΔH for the dissolution of θ-phase plotted as a function of homogenisation time and homogenisation temperature.

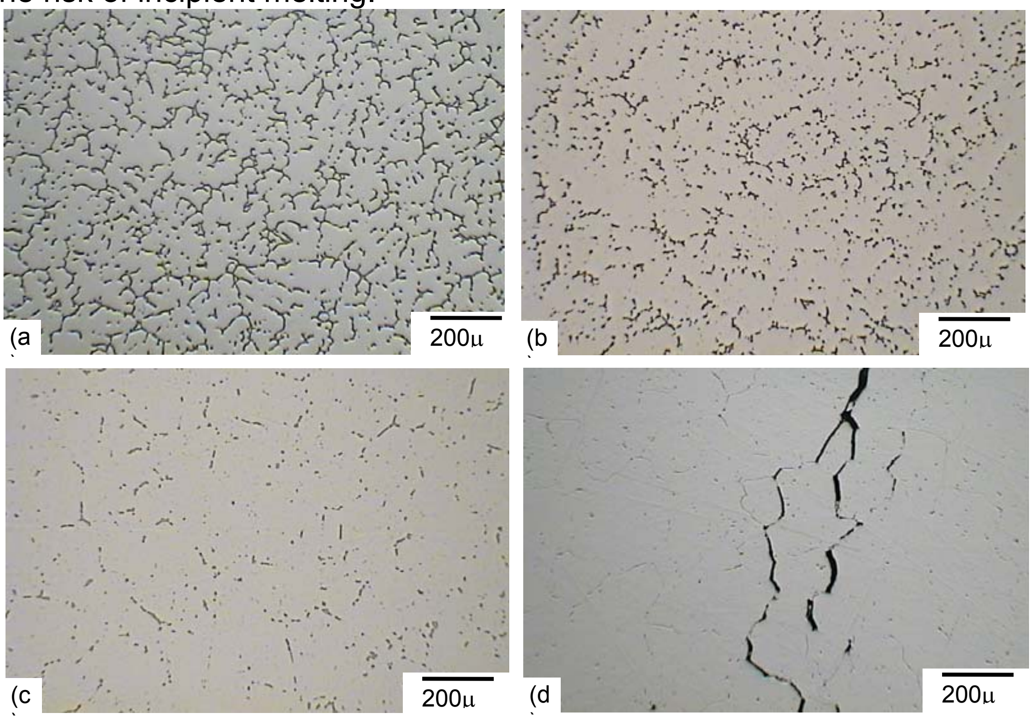

Figure 5: Optical microstructure of aluminium alloy AA2001 in (a) the as-cast condition, and after various homogenisation treatments: (b) 4h at 400°C, (c) 4h at 525°C and (d) 4h at 545°C.

3.3 Influence of Cooling Rate

Figure 6 shows a plot of storage time versus hardness for the four different homogenisation practices over a two week period. It can be observed that none of the samples which cooled directly or slow furnace cooled achieved a hardness level suitable for extrusion. The hardness level of the sample homogenised at 525°C, and directly cooled (A-DC) increased from 68HV10 to 96HV10 after only four days storage and further increased to 103HV10 over the remainder of the two week period, i.e. to a hardness level higher than that observed in the alloy in the as-cast condition. However a hardness increase from only ≈65HV10 to 74 HV10 was observed for the sample homogenised at 525°C and furnace cooled (A-FC), thus indicating that the alloys susceptibility to natural ageing decreases with cooling rate. It should further be noted that the hardness level of the sample homogenised at 400°C and directly cooled (B-DC) only increased from 66HV10 to 70HV10 over the entire two week period, even thought it was subject to a direct cool as opposed to a furnace cool, thus indicating that the alloys susceptibility to natural ageing also decreases with a lower homogenisation temperature. Both these observations are further confirmed by the 400°C step furnace cooled sample (B-SFC), which showed a hardness level below 65HV10 over the two week period, which was then acceptable for extrusion. This pattern of results are confirmed by the equivalent stress-strain curves (Figure 7) which indicate an equivalent stress of 375Mpa, 290MPa and 240Mpa, at a strain of 0.5, for as-cast, A-FC (525°C, furnace-cooled) and B-SFC (400°C, step-furnace-cooled) samples respectively, after a two week storage period.

From DSC and optical image analysis it is known that at the homogenisation temperature of 525°C (practice A) most of the eutectic phase is placed into solid solution. It is therefore unlikely that subsequent direct cooling (DC) allows sufficient time or driving force for the solid solution to re-precipitate into the desired microstructure i.e. an intradendritic distribution of fairly large particles [9]. Thus a supersaturated solid solution is obtained, resulting in a high initial hardness due to solid solution strengthening, and moreover providing a microstructure highly susceptible to natural age hardening. It therefore follows that samples homogenised using practice B are less susceptible to age hardening since there is less eutectic initially placed into solution at 400°C. Furthermore practice B combined with step-cooling (B-SFC) provides a sample with a microstructure suitable for extrusion that is not susceptible to natural ageing, since the step-cool acts to promote the re-precipitation of the solid solution into the desired microstructure. Whilst practice B-SFC is acceptable, it can not be considered commercially, as it does not utilise the high diffusion rate attained at the maximum homogenisation temperature of 525°C. The step furnace cool is also of a duration that is not viable (Figure 3).

By considering the experimental results, a homogenisation practice consisting of a ramp time of 2h, a dwell of 4h at 525°C and a shortened stepped-furnace-cool (Figure 3) based on C-curve data for AA2017 [6] was developed (A-SFC). As a result a sample with a hardness level <65HV10, remained suitable for extrusion over the two-week period. An intradendritic distribution of fairly large particles can clearly be seen in the final microstructure (Figure 8a). This is in contrast to that obtained after a direct cool from 525°C, where the majority of the solute is in either solid solution, or the early stages of precipitate hardening (Figure 8b).

3.4 Extrusion and Ageing

The extrusion trial was successful with all 70 billets extruded free from defects. It was observed that the final proof stress and UTS achieved in the T6 condition after homogenisation practice A-SFC and B-SFC were similar (Table 2). Moreover the final properties of the components met those required for service.

Table 2: Final mechanical properties as a function of the Homogenisation Practice.

Homogenisation Practice	Aged Condition	0.2% Proof (Mpa)	UTS (Mpa)	Elongation (%)
A-SFC	T6	396	464	12.0
B-SFC	T6	393	460	11.8

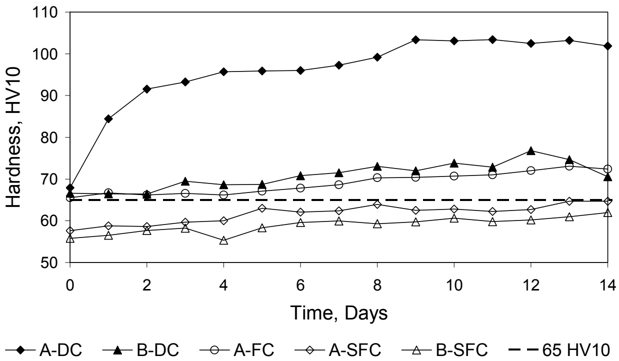

Figure 6: Workability (hardness HV10) observed over a two week period for different homogenisation and cooling practices.

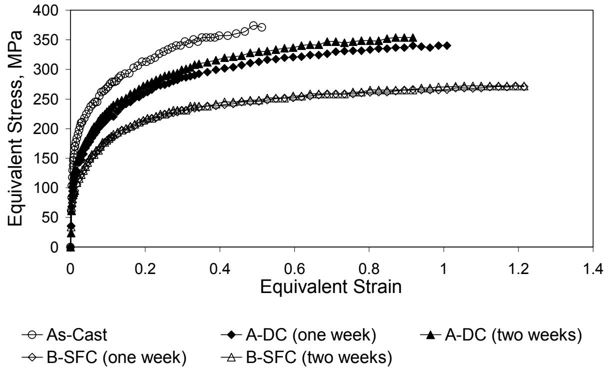

Figure 7: Equivalent stress-strain curves for two different homogenisation and cooling practices, observed at one week intervals over a two week period.

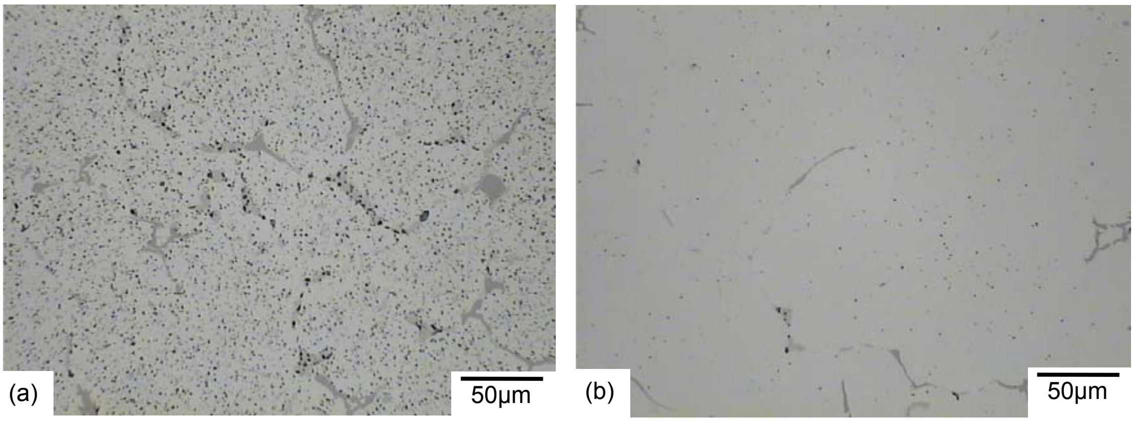

Figure 8: Resulting microstructures after a 4h saturation treatment at 525°C and (a) step furnace cool to ambient (A-SFC) and (b) direct cool to ambient (A-DC).

4. Conclusions

A commercially feasible practice for the homogenisation of 2xxx series aluminium alloy billets for the production of light-weight defect-free components has been successfully developed by a systematic study, using advanced material characterisation techniques. It was observed that the initial level of workability after homogenisation and subsequent susceptibility to natural ageing during storage is dependent on the amount of solute in solution, as governed by both the homogenisation temperature and cooling rate. The condition of the alloy may be monitored by hardness. For defect free extruded material HV10 should be $\leq$65.

A step furnace cool proved most effective in obtaining the desired microstructure for extrusion, which consisted of an intradendritic distribution of fairly coarse $CuAl_2$ particles.

Acknowledgements

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