Effects of process variables on the mechanical and physical properties of an Al–Cu–Mg powder metallurgy alloy

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Abstract

The objective of this research was to analyze the effect sizing and heat treatment had on the properties of a 2xxx series aluminum powder metallurgy alloy. Al–2.3Cu–1.5Mg–0.5Sn was used for experimental work. Natural and artificial aging was studied, as well as various sizing pressures and delay times between process steps (ex. sintering and sizing). It was determined that sizing increased tensile yield strength but decreased tensile ductility and fatigue resistance. Yield strength increased as much as 32% when sized at 400 MPa, whereas ductility and fatigue strength were reduced by up to 51% and 31%, respectively. Delay time between sintering and sizing (T2 samples) did not appear to significantly impact mechanical properties. Through differential scanning calorimetry it was determined that sizing impacted the precipitation of S-type precipitates. It was inferred that there was a preferential tendency for S1-type precipitates to form in sized samples, suppressing full aging to S2-type.

Keywords Aluminum powder metallurgy · Al–Cu–Mg · Sizing · Fatigue · Tensile · DSC

1 Introduction

Aluminum powder metallurgy (PM) alloys have recently been under intense study due to a push from automotive manufacturers to facilitate reductions in vehicle weight and, in turn, enhance fuel economy. PM is an attractive manufacturing method due to the high volume production capabilities and the reduced need for machining of parts. In particular, 2xxx series aluminum alloys have proven to be viable candidates for automotive PM components due to their effective compaction and sintering responses. The typical processing route for press-and-sinter aluminum PM products includes powder blending, compaction, sintering, sizing, heat treatment and other secondary operations as necessary. A significant amount of recent work has been done on the processing of 2xxx series alloys. For example, Cooke et al. [1] examined the impact of lower Cu/Mg ratios and Schaffer et al. [2] explored the effect of tin and nitrogen on sintering. Other

work has studied the industrial processing of Al–Cu–Mg alloys [3] or the effect of hot forging on the 2xxx series [4].

Although many of these studies have comprehensively examined various processing methods and the effect of different variables, the impact of sizing has been largely neglected. Specifically, how sizing in combination with various heat treatments affects the mechanical and physical properties of Al–Cu–Mg alloys. This information could be crucial to industrial manufacturers because sizing is essential in commercial aluminum PM production to control final dimensional tolerances. Automotive manufacturers are particularly interested in the impact of sizing on fatigue properties since components can undergo millions of loading cycles in their lifetime. Due to the microstructure (especially porosity) of PM parts, consistent production methods are necessary to produce consistent fatigue results. Grayson et al. [5] discovered that fatigue cracks in Ampalloy 2712 (Al–3.8Cu–1Mg–0.7Si–0.1Sn) initiated in pores located just below the surface of the material. This

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highlights the importance of reducing porosity in aluminum PM components.

Sizing is known to change the tensile properties of aluminum PM alloys [6]. Specifically, sizing has produced gains in yield strength (YS) and ultimate tensile strength (UTS) while decreasing ductility in an Al–4.5Cu–1.5Mg PM alloy. The gains in YS and UTS were attributed to strain hardening in the material as well as increasing the refinement of S-type precipitates (as discussed in the following paragraph). Although the scope of that study did not include fatigue testing, more recent work on a 7xxx series aluminum alloy has demonstrated significant variations in fatigue strength by altering the sizing parameters [7]. Here, the three-point bending fatigue strength of an Al–5.6Zn–2.5Mg–1.6Cu PM alloy was reduced by 23% when sizing was done immediately after quenching. The authors attributed this decrease in fatigue strength to an increased amount of incoherent η precipitates within the sized microstructures. Here, sizing had altered the solid-state reactions that occurred during precipitation hardening and ultimately catalyzed the formation of η precipitates.

2xxx series aluminum alloys are principally strengthened by precipitates. The type and amount of precipitates present in alloys depends on the alloy chemistry and processing route (e.g. heat treatment method). Various studies, have determined that S-type (Al_2CuMg) precipitates are prevalent in aluminum alloys with low Cu/Mg ratios [8]. θ -type (CuAl_2) precipitates are also common in Al–Cu–Mg alloys, but are more dominant in alloys with a higher Cu/Mg ratio. Cooke et al. [1] found that good tensile properties could be produced by promoting S-type precipitates through a low Cu/Mg ratio in certain PM alloys. Cold working (sizing) is expected to enhance and refine S-type precipitation because this phase is known to nucleate on dislocations [9]. In the present work, an alloy with the composition of Al–2.3Cu–1.5Mg–0.5Sn was studied. This composition equates to a low Cu/Mg ratio of 1.53/1. Hence, S-type precipitates should be the main strengthening mechanism present.

There are various arguments on the how the precipitation sequence of the S phase proceeds. Early works [10] stated the sequence as:

Since then, the precipitation sequence has been studied in more detail [11]. Here, it has been proposed that 2 different S-type phases are present and suggested the following precipitation sequence:

The S1 type precipitates are metastable and can transform into S2 precipitates depending on processing parameters such as aging time. The precipitation sequence described in [R2] is generally agreed upon and various researchers have confirmed the presence of S1 and S2 precipitates and that there is a distinct crystallographic difference between them [12]. Given the complexities inherent to precipitation hardening of 2xxx series alloys and the documented influence of cold work/sizing on other aluminum PM alloys, work was completed in this study to determine if similar transitions occurred in a PM alloy of low Cu:Mg ratio and how this impacted fundamental mechanical properties.

2 Materials

The PM alloy used in this work was a 2xxx series aluminum alloy with a nominal composition of Al–2.3Cu–1.5Mg–0.5Sn (wt%). The starting powder blend was comprised of air atomized elemental aluminum from Ecka Granules (99.8% purity), a 50:50 Al–Cu master alloy (99.5% purity), and atomized elemental magnesium (99.8% purity) and tin (99.9% purity) powders. The particle size distribution of each powder was measured using laser light scattering (Malvern Mastersizer 3000). A summary of the nominal particle sizes for each powder is presented in Table 1. In addition, 1.5 wt% Licowax C was added to each mixture of the alloy to facilitate die compaction.

3 Experimental techniques

Blending of the metallic powders was done with a Turbula mixer in the following order: elemental aluminum was blended with the Al–Cu master alloy for 40 min, magnesium and tin were then added and blended for 45 min, and finally, Licowax C was added and blended for an additional 55 min. The powder blend was then die compacted in a floating die using an Instron 5594-200HVL load frame with a 1 MN load capacity. All samples were compacted at 200 MPa to form the required green part and remain consistent with typical industrial processing conditions

Table 1 Nominal particle sizes (microns) of the metallic powders utilized

Powder	D ₉₀	D ₅₀	D ₁₀
Aluminum	219	116	63
50:50 Al–Cu master alloy	76	36	13
Magnesium	48	32	14
Tin	36	16	6

employed for the powder system of interest. The densities of green compacts were 2.54 g/cm³. Bars of two geometries were compacted in this fashion—transverse rupture strength (TRS) bars (31.7 × 12.7 × 10 mm) and charpy bars (75 × 12.7 × 12.7 mm). Green samples were then sintered in a three-zone tube furnace under a flowing nitrogen atmosphere (99.999%). The tube furnace was evacuated and back-filled with nitrogen twice before sintering to minimize the level of oxygen in the chamber. The thermal profile of the sintering process consisted of a 20-min hold at 400 °C for de-lubrication and a 20-min hold at 630 °C for sintering. Samples were then cooled to ambient within a water jacketed section of the furnace under flowing nitrogen. The temperature profile was recorded for each sinter by means of a type K thermocouple clamped on the sintering tray.

After sintering, samples were subjected to heat treatment and sizing. T1, T2, T3, T4, T6, and T8 treatments were all utilized. Table 2 describes the heat treatment and sizing processes applied in each instance.

Sizing of the sintered samples was completed using the Instron 5594-200HVL load frame at pressures ranging from 200 to 450 MPa. This targeted an OAL reduction of up to ~5.5% at 450 MPa. To investigate the impact of delay time between sintering and sizing on the alloy, T2 samples were naturally aged for 1, 10, 100, and 1000 h before sizing at 450 MPa. Densities were measured using an Archimedes approach based on oil infiltration as described in Metal Powder Industries Federation (MPIF) standard 42 [13]. The density of as-sintered specimens was 2.71 g/cm³ (99.4% of full theoretical). Sizing at 200 MPa manifested no measurable increase in this attribute. Sizing at the highest pressure considered (450 MPa) prompted a minor gain of only

0.01 g/cm³. As such, it was confirmed that sizing had no significant impact on the final density of the specimens.

Tensile data were obtained from machined threaded cylindrical tensile samples with a diameter of 5.00 mm (±0.05). The same Instron frame used for compaction and sizing was utilized for tensile testing, but when equipped with a 50 kN load cell rather than the 1 MN load cell. An Epsilon 3542 extensometer was used to measure engineering strain during testing. Samples were loaded at a rate of 2 MPa/s until fracture. Three samples were tested for each processing parameter and the average reported. A three-point bend fixture was used together with an Instron Model 1332 servo-hydraulic frame operated at 25 Hz for fatigue testing. The fixture had a 24.7 mm span between the lower pins that were made from hardened tool steel and had an outer diameter of 3.2 mm. TRS bars were tested in this fixture in accordance with the staircase method (runout of 1 million cycles; 5 MPa step size) to determine fatigue strength as described in MPIF Standard 56 [13]. TRS bars were pre-loaded to 0.1 kN and an R ratio of 0.1 was utilized. All TRS samples were de-burred prior to testing by light sanding on 600 grit SiC paper. No surface machining was carried out, to ensure that the sintered/sized surfaces were tested directly. A total of 12–15 samples were tested when evaluating fatigue properties for each processing sequence considered. Optical micrographs of the unetched microstructures were obtained using a Zeiss Axio optical microscope. Here, samples were sectioned, mounted in bakelite and then ground/polished through a standard sequence of silicon carbide grinding pads, diamond suspensions, and colloidal silica using a Struers Tegramin auto-polisher. Differential scanning calorimetry (DSC) was used to study heat treatment

Table 2 Descriptions of the tempers applied

Temper	Description
T1	1. Naturally aged to stable condition at 21 °C (minimum of 1 week)
T2	1. Sized at 200 or 450 MPa 1 h* after sintering 2. Naturally aged to stable condition (minimum of 1 week)
T4	1. Solutionized @ 530 °C for 2 h 2. Water quench 3. Naturally aged to stable condition at 21 °C (minimum of 1 week)
T3	1. Solutionized @ 530 °C for 2 h 2. Water quench 3. Sized at 450 MPa 1 h after quenching 4. Naturally aged to stable condition at 21 °C (minimum of 1 week)
T6	1. Solutionized @ 530 °C for 2 h 2. Water quench 3. Artificially aged @ 200 °C for 10 h
T8	1. Solutionized @ 530 °C for 2 h 2. Water quench 3. Sized at 450 MPa either 1 or 24 h after quenching 4. Artificially aged @ 200 °C for 10 h

*The impact of various delay times were also investigated for the T2 condition

and precipitate development. A TA Instruments SDT Q600 V8.1 was used for these experiments. DSC samples were prepared by machining small cylindrical samples with a nominal diameter and height of 4 mm from TRS bars. Samples were heated in alumina crucibles from 50 to 550 °C at a rate of 10 °C/min in air. A pure aluminum sample was initially tested and the resultant mass normalized heat flow was subtracted from the mass normalized trace acquired from each PM sample to isolate the thermal events related to precipitation-based reactions. Replicate samples were tested in each instance to confirm that a particular heat flow response was reproducible.

4 Results and discussion

4.1 Naturally aged tempers (T1 and T2)

To test the impact that sizing had on the as-sintered alloy, mechanical testing was carried out on sintered (T1) and sintered + sized (T2) samples. Two process variables were explored: sizing pressure and the delay time between sintering and sizing. Two sizing pressures were tested that embody the range frequently encountered in commercial practice. Similarly, as sintered parts may sit for extended periods of time before they are sized, various delay times were tested. This would allow the sintered specimen to naturally harden to different extents prior to cold working (sizing), and therefore could change how sizing impacts the mechanical properties.

Figures 1, 2, 3 and 4 show the yield strength (YS) and ductility of the T1 and T2 samples. From these results, it was clear that increased sizing pressure led to higher YS, but lowered ductility at highest sizing pressure considered. The impact of delay time between sintering and sizing was not as pronounced. Here, minimal changes

Fig. 1 Impact of increasing sizing pressure on the yield strength of T1/T2 tempers. Average values indicated in bold italicized text. Sized samples have darker shading

Fig. 2 Impact of time delay between sintering and sizing on the yield strength of the T2 temper. Average values indicated in bold italicized text. All specimens sized at a pressure of 450 MPa

in YS and ductility were observed even when delay time was increased to 1000 h. With experimental error taken into consideration, these transitions were viewed to be limited. Overall, sizing pressure was the more influential variable as it provided tangible gains in YS and modest reductions in ductility within T2 specimens.

It was determined that sizing also had an impact on fatigue properties. As with the tensile tests, the impact of sizing pressure and delay time were again investigated. Figures 5 and 6 show the average fatigue strengths measured for the T1 and T2 samples. In all fatigue charts shown in this paper, the upper and lower ends of the error bars represent the $\sigma_{10\%}$ and $\sigma_{90\%}$ fatigue stress (stresses at which 10% and 90% of the samples will pass) respectively.

As sizing pressure increased, fatigue strength was negatively impacted. For example, when sized at 450 MPa, the fatigue strength of the alloy decreased by 16% when compared to the unsized (T1) processing route. This came as somewhat of a surprising result, considering the positive impact sizing had on yield strength. Figure 6 shows that increasing the delay time between sintering and sizing did not seem to have any significant impact on fatigue strength.

To determine why sizing negatively affected the fatigue properties of the alloy, microstructural analyses were completed. Optical micrographs of the unetched microstructures are shown in Fig. 7a–f. No significant differences in the microstructures were found that would signal why sizing had changed the mechanical properties. The images showed some residual porosity and the presence of secondary phases in all instances, but no cracks or other obvious abnormalities could be seen at any magnification. SEM analyses resulted in the same findings. Hence, it was hypothesized that the changes in mechanical properties arose due to variation(s) in precipitate development which

Fig. 3 Impact of increasing sizing pressure on the ductility of T1 and T2 tempers. Average values indicated in bold italicized text. Sized samples have darker shading

Fig. 4 Impact of time delay between sintering and sizing on the ductility of T2 temper. Average values indicated in bold italicized text

could not be observed through either of the imaging techniques applied.

To gain insight into how precipitate development may have been affected by sizing, DSC trials were run on each of the T1 and T2 samples processed. DSC is an insightful technique known to yield critical data on the precipitates

present in aluminum alloys and transformation therein [14]. Those pertaining to the effects of sizing pressure are shown in Fig. 8. All traces contained peaks of an exothermic and endothermic nature that varied in position and magnitude as sizing was applied. As each peak corresponds to an event in the precipitation sequence of the S phase [11], it was concluded that sizing impacted precipitation hardening. The first endothermic effect occurred at $\sim 230^\circ\text{C}$ and was believed to be the dissolution Cu–Mg clusters and possibly S'. This was followed by a relatively large exothermic peak ascribed to be the formation of the two variants of the S-phase precipitates; namely, S1 and/or S2. This peak occurred over a broad temperature range of $240\text{--}410^\circ\text{C}$ in the T1 sample. In sized specimens this peak shifted to lower temperatures, became narrower, and more intense. It was also noted that these transitions became more acute as sizing pressure increased. These changes were believed to be due to the facts that S-type precipitates are known to preferentially nucleate on dislocations [9] and that the strain hardening instilled through sizing would introduce progressively higher concentrations of these defects as the applied pressure increased. This would have provided progressively higher concentrations of nucleation sites to

Fig. 5 Impact of increased sizing pressure on the fatigue strength of T1 and T2 tempered specimens. Average values indicated in bold italicized text. Sized samples have darker shading

Fig. 6 Impact of time delay between sintering and sizing on the fatigue strength of specimens in the T2 temper. Average values indicated in bold italicized text

facilitate precipitate formation, thereby shifting the peak to lower temperatures. It is also postulated that the width of the peak narrowed due to a preferential tendency to form S1 precipitates. This point is discussed in more detail at the end of the results section. Finally, there was a broad endothermic peak. Commonly denoted as a dissolution trough, this transition corresponded to the dissolution of S1 and/or S2 phases.

Additional DSC trials were then completed on T2 samples that were naturally aged for times of 1 h and 1000 h prior to sizing. The resultant heat flow data are shown in Fig. 9. Both curves were in strong agreement, indicating that the S phase precipitate development was quite comparable in these specimens. This was consistent with the fact that the delay time between sintering and sizing did not impart any significant transitions in mechanical properties.

To gain an improved understanding of precipitate transitions during natural aging, a sintered specimen was solutionized at 530 °C for 2 h, water quenched, transferred directly to the DSC and heated. The resultant heat flow data are presented in Fig. 10 along with the trace acquired from the T1 sample previously shown in Fig. 8. A notable distinction was that data from the quenched specimen exhibited an exothermic peak at ~ 110 °C. This was believed to coincide with the first step in the S1/S2 reaction sequence—Cu–Mg cluster formation. This peak was not observed in any of the T1/T2 samples indicating that the bulk of the precipitates in these specimens had already evolved to (or beyond) the cluster formation stage prior to heating in the DSC; in effect, during the natural age hardening period applied. Subsequent peaks coinciding with cluster dissolution, S1/S2 precipitation, and finally S1/S2 dissolution were remarkably like those generated from the T1 specimen in terms of their respective widths and temperatures of occurrence. As none of these peaks

were shifted towards lower temperatures, this further substantiated the notion that sizing was in fact the driving force behind the transitions observed in Fig. 8.

4.2 Artificially aged tempers (T6 and T8)

Solutionizing, quenching and artificially aging is commonly used in industry since it results in higher tensile strengths for 2xxx series aluminum PM alloys. For this reason, it was necessary to investigate if sizing had a comparable impact on T6/T8 tempered specimens of the alloy (i.e. positive on yield strength, yet negative on fatigue) relative to T1/T2 counterparts. Two variations on the T8 temper were investigated: one with a 1 h delay between solutionizing and sizing, and another with a 24 h delay between solutionizing and sizing. All T8 samples were sized at 450 MPa (~ 5.5% reduction) as this pressure was aligned with commercial practice and was found to have a prolific effect in accordance with T1/T2 data. Data on yield strength and ductility for the T6 and T8 samples are shown in Figs. 11 and 12 respectively.

Sizing increased the yield strength of the alloy by ~ 79 MPa which equated to a net gain of 32% relative to the T6 counterpart. Conversely, a statistically relevant trend was not discernable when ductility data were assessed as all average values fell within a relatively narrow range of 3–5%. The delay time between sintering and sizing failed to bear an obvious impact on either property suggesting that it was a factor of lesser influence. Relative to the T1/T2 specimens, those processed into the T6/T8 states offered enhanced yield strength at the expense of reduced tensile ductility.

Interestingly, sizing was again found to have a negative impact on the fatigue strength as shown in Fig. 13. Here, the T8 tempered products exhibited decreases in fatigue strength on the order of 25–30% when compared to those in the T6 temper. Although there may have been a slight increase in fatigue strength when the T8 sample was aged 24 h before sizing, additional testing would be necessary to determine if this was a statistically meaningful change.

As precipitate differences were found to underpin mechanical property differences in T1/T2 samples, DSC was again implemented to gain comparable insight on T6/T8 materials. Samples of T6, T8 (1 h), and T8 (24 h) were all run in the DSC as shown in Fig. 14.

The T6 temper exhibited heat flow characteristics similar to those observed in the T1 temper in that peaks believed to coincide with cluster dissolution, S1/S2 precipitation, and S1/S2 dissolution were all observed. The largest peak was due to S1/S2 precipitation and occurred at 310 °C. Interestingly, this peak was significantly diminished in both T8 samples, thereby indicating that most of the precipitates present in the T8 samples had already

Fig. 7 Optical images of the T1/T2 variants considered. **a** No sizing, **b** sized at 200 MPa, **c** sized at 450 MPa, **d** 10 h delay time prior to sizing, **e** 100 h delay time prior to sizing, **f** 1000 h delay time prior to sizing

aged to the S1 or S2 state before heating in the DSC. This implied that precipitation reactions occurred more readily in the sized T8 samples. This was consistent with T1/T2 data in that sizing shifted the S1/S2 peak to significantly lower temperatures with sizing (Fig. 8). Hence, it was evident that sizing had once again imparted a direct impact on precipitation and was therefore the underlying factor in the observed transitions in yield strength and fatigue resistance.

4.3 Solutionized, quenched, and naturally aged tempers (T3 and T4)

The final processing routes explored in this study were the T3 and T4 tempers. Here, select samples were solutionized at 530 °C, quenched in room temperature water, and then naturally aged to a stable state to impart a T4 temper. The T3 temper was achieved by cold working (sizing) the samples immediately after quenching. Tensile data acquired

Fig. 8 DSC traces acquired from specimens in the T1 and T2 tempers

Fig. 9 DSC traces acquired from samples in the T2 temper sized 1 h and 1000 h after sintering. Both specimens sized at 450 MPa

Fig. 10 Comparison of the DSC traces acquired from a T1 sample and one that was heated immediately after being solutionized and water quenched

Fig. 11 Comparison of the tensile yield strength measured for specimens in the T6 and T8(x) tempers. Average values indicated in bold italicized text. Sized samples have darker shading

Fig. 12 Comparison of the tensile ductility measured for specimens in the T6 and T8(x) tempers. Average values indicated in bold italicized text. Sized samples have darker shading

Fig. 13 Comparison of the fatigue strength measured for specimens in the T6 and T8(x) tempers. Average values indicated in bold italicized text. Sized samples have darker shading

from specimens in these tempers are shown in Figs. 15 and 16.

As expected, the sized (T3) samples exhibited a higher yield strength than those in the T4 state. This increase in yield strength came at the expense of ductility, which was halved in the sized temper. Overall, the T3/T4 samples displayed lower yield strengths, but higher ductilities than the T6/T8 samples. As with the other processing routes, fatigue strength was again negatively affected by sizing. Figure 17 shows that fatigue strength decreased 29% when the T3 temper was utilized. Interestingly, the T4 temper specimens displayed the highest fatigue strength of any processing route. These results align with what is commonly seen in wrought aluminum alloys in that T4 samples tend to have higher fatigue strengths and ductilities, whereas T6 samples tend to have higher yield and tensile strengths [15].

Figure 18 shows the DSC results acquired from the T3 and T4 tempered materials. As with the other processing routes, the sized (T3) temper displayed a different heat flow trace than the its non-sized counterpart. In particular, the principal exothermic peak of the trace was shifted to a lower temperature, was narrower, and more intense. As previously mentioned, all such transitions implied an intensified concentration of S1 precipitates nucleating on sizing-induced dislocations.

Table 3 shows an overall summary of averaged mechanical properties for all tempers explored in this study. In the case where multiple variations of the temper were assessed (T2 and T8), only the variation with the highest fatigue strength is shown. Considering all the different tempers, sizing consistently enhanced yield strength but negatively impacted fatigue response, and in general, tensile ductility.

Based on the DSC results for all tempers, it was quite apparent that sizing had impacted precipitate development within the alloy. The DSC traces for all sized tempers (T2, T3, and T8) were clearly distinguishable from the as-processed tempers (T1, T4, and T6). The principal difference in each case was related to the temperature at which the main exothermic peak (ascribed to S1/S2 formation) occurred and its width. Table 4 summarizes these transitions observed in this peak as well as the peak width. For the T1/T2 and T3/T4 tempers (Figs. 8, 18), DSC findings indicated that sizing consistently shifted this peak to a lower temperature, increased its magnitude, and decreased the peak width to a smaller temperature range. Similar peak shifts were seen in AA2324 cold worked samples [11]. Here it was concluded that in solution treated, quenched and subsequently cold worked samples, extensive formation of S1 occurs, whilst the formation of S2 was suppressed. Other studies also observed this peak shift under comparable circumstances [16]. Interestingly, these researchers also

Fig. 14 DSC heat flow traces acquired from T6 and T8(x) tempered specimens

Fig. 15 Comparison of the yield strength measured for specimens in the T4 and T3 tempers. Average values indicated in bold italicized text. Sized sample has darker shading

Fig. 16 Comparison of the ductility measured for specimens in the T4 and T3 tempers. Average values indicated in bold italicized text. Sized sample has darker shading

Fig. 17 Comparison of the fatigue strength measured for specimens in the T4 and T3 tempers. Average values indicated in bold italicized text. Sized sample has darker shading

observed a small exothermic effect on the shoulder of the S formation peak. They believed this to be associated with the formation of the S2 type precipitates. This shoulder effect was not observed in the T1/T2 and T3/T4 DSC results of this study because according to Wang and Starink [11] only strain energy and interfacial energy are released during the transformation from S1 to S2, meaning that the expected heat effect associated with the transformation would be relatively small. This small heat effect from the transformation of S1 to S2 could be masked by the much larger heat effect of the S1 phase that forms at ~300 °C in the samples that were not sized. Hence, although samples in the T1, T4, and T6 temper showed no signs of this shoulder effect, the S2 type precipitates were still believed to be present in these tempers. Conversely, it is proposed that the sized tempers likely have a higher concentration of the S1 precipitates which prompted a negative impact on fatigue resistance.

Fig. 18 DSC heat flow traces acquired from T4 and T3 tempered specimens **Table 3** Summary of the average mechanical properties measured for the Al–2.3Cu–1.5Mg–0.5Sn PM alloy in the different tempers considered

	Yield strength (MPa)	Max elongation (%)	Fatigue strength $\sigma_{50\%}$ (MPa)
T1	185	10.9	170
T2	242	8.8	144
T4	196	13.6	200
T3	235	6.6	142
T6	246	4.4	158
T8	324	3.1	109

Table 4 Effect of sizing on the peak ascribed to S1/S2 precipitation

Temper	Peak temperature		Peak width	
	No sizing (°C)	Sized (°C)	No sizing (°C)	Sized (°C)
T1/T2*	330	270	160	70
T6/T8**	310	240	100	80
T4/T3	300	270	120	70

* Value for sample sized at 450 MPa

**Value for sample with 1 h delay between solutionizing and sizing

Interestingly, the S1/S2 peak was effectively absent in DSC data from the T8 temper, which indicates that all the precipitates were already pre-aged to the S1/S2 state. This thereby implied that it was particularly easy to form S phase precipitates in this manner of heat treatment. The data given in Table 4 for the sized T8 sample is much more subjective than for the other tempers due to the nature of the S1/S2 peak in this case. Although not directly seen in the DSC trace, it's likely that the T8 temper, in much the same way as the T2 and T3 tempers, inhibits the formation

of S2 precipitates and would therefore have a higher amount of the S1-type. Overall, the DSC traces for the T8 samples were significantly different from the T6 counterparts again indicating that sizing impacted the precipitate development.

5 Conclusions

In assessing the effects of sizing and heat treatment process on the PM alloy (Al–2.3Cu–1.5Mg–0.5Sn) of interest the following conclusions were reached:

1. Sizing lead to meaningful increases in the tensile yield strength in all tempers, but this was accompanied by appreciable losses in fatigue resistance and tensile ductility.
2. The mechanical property transitions between unsized and sized specimens were found to scale with the applied sizing pressure.
3. The delay time between sintering and sizing did not impact the mechanical properties significantly.
4. DSC assessments were consistently different for sized specimens (regardless of temper). In particular, the principal exothermic peak believed to be due to S1/S2 precipitation was shifted to lower temperatures, became narrower, and more intense. It was inferred that this signified a preferential tendency for the formation of S1-type precipitates in the sized specimens.

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Compliance with ethical standards

Conflict of interest The authors declare that they have no conflict of interest.

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