Effect of Ag and Cu addition on the strength and electrical conductivity of Al-Mg-Si alloys using conventional and modified thermomechanical treatments

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Abstract

The effect of adding Ag and Cu on the strength and electrical conductivity of Al-Mg-Si conductor alloys was investigated using conventional and modified thermomechanical treatments. In the conventional thermomechanical treatment, the addition of Ag and Cu moderately increased the strength from 296 MPa to 305-316 MPa above the minimum required electrical conductivity (52.5% IACS) compared to that of the base alloy. However, the modification of the thermomechanical treatment could maximize the efficiency of Ag and Cu addition at strengths above 52.5% IACS, exhibiting that the strength was increased from 317 MPa to 341-348 MPa with the Ag and Cu additions. All alloys under the modified thermomechanical treatment (MTMT) showed a superior strength and electrical conductivity compared to their counterparts under the conventional thermomechanical treatment (CTMT), resulting in a wider window in the high end of strength and electrical conductivity. In addition, MTMT led to a shorter post-aging time to reach the minimum required EC, compared to CTMT. The precipitate characteristics under both thermomechanical treatment conditions were analyzed and quantified using differential scanning calorimetry and transmission electron microscopy (TEM). The TEM results revealed that the alloys under MTMT had a higher precipitate number density by more than three times compared to their counterparts under CTMT, leading to the higher strength levels in the MTMT alloys. Strength and electrical resistivity models were then applied to understand the multiple contributions of the main strengthening mechanisms and microstructure features to the mechanical strength and electrical conductivity.

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1. Introduction

Aluminum conductor alloys are increasingly used in high voltage electrical transmission lines. In general, the electrical conductivity of aluminum is approximately 62% that of copper. However, considering the equal current-carrying capacity, the weight of the aluminum conductor is only half that of the copper conductor due to its lower density [1]. Among aluminum conductors, heat-treatable Al-Mg-Si conductor alloys are the widely used conductive materials in power transmission due to their excellent combination of high strength with the desired electrical properties in recent decades, strengthened by nanosized precipitates via aging treatment [2-4].

The conventional thermomechanical treatment (CTMT) in the fabrication of Al-Mg-Si conductors is comprised of quenching from the solutionization temperature and cold wire drawing followed by post-aging treatment [3]. However, some studies were allocated to the modified thermomechanical treatment (MTMT) to attain a better optimization of the strength and electrical conductivity (EC). It might consist of quenching from the solutionization temperature, pre-aging, wire drawing, and post-aging treatment [5-7]. Lin et al. [5] suggested an MTMT, in which pre-aging for 2 h at 180 °C led to a better trade-off between strength (301 MPa) and EC (58.9% IACS) for the Al-0.62Mg-0.57Si alloy. Liu et al. [6] also proposed pre-aging at 180 °C for 5 h for an Al-1Mg-0.5Si-0.8Cu alloy, resulting in a higher strength and EC compared to CTMT. In our previous research, it was shown that a pre-aging treatment (5 h at 180 °C) in MTMT for the Al-0.62Mg-0.62Si alloy could remarkably increase the precipitate number density at the expense of the dislocation number density. Therefore, the superior precipitate strengthening overcame the inferior strain hardening, leading to a better compromise between strength (~369 MPa) and EC (~53% IACS) [8].

The impacts of principal alloying elements (Mg and Si) and additional alloying elements (Fe, La, Ce, and Sr) on the strength and electrical conductivity in 6xxx series conductors have been the subject of multiple studies [9-16]. Regarding the principal alloying elements (Mg and Si), an alloy with an Mg/Si of ~ 1 showed a better combination of strength and EC, provided the strength is taken as a main concern [9-11]. Excessive magnesium (Mg/Si > 2) exhibited an almost insignificant improvement in the strength and EC [9], while excessive silicon (Mg/Si < 1) gave a limited window of strength and EC (above the minimum required EC, 52.5% IACS for AA6201 conductor alloys [17]) with a lower strength compared to the alloy with an Mg/Si of ~ 1 [10, 11]. The addition of Fe, La, and Ce to Al-Mg-Si conductor alloys led to an improvement in the electrical conductivity at the expense of strength [12-14]. In contrast, the Sr addition could result in a better optimization of strength and EC in Al-0.6Mg-0.6Si conductor alloys due to a higher precipitation degree and intermetallic modification [15, 16].

Copper and silver additions have recently been taken into account in Al-Mg-Si alloys to promote the age hardening response [18]. It was reported that Mg-Cu and Mg-Ag clusters could form in the early stage of aging due to the stronger interaction energy between Cu/Ag atoms and Mg atoms. Therefore, these clusters with Mg-Si clusters could act as nucleation sites for the subsequent precipitates and enhance precipitation strengthening [18-20]. It was determined that Cu-/Ag-added alloys exhibited a higher strength than the base alloy due to the higher precipitate number density [19, 20]. The Cu addition could lead to the formation of L and Q' precipitates, which coexist with β'' phases in the peak-aged condition [20, 21]. It was reported that Cu additions (0.2-1.1) wt.% to 6xxx series aluminum alloys showed a significant superiority in terms of strength [18-23]. Duan et al. [23] reported that Cu-added alloys (0.3-0.9 wt.% Cu) showed an inferior EC in the early aging time (up to 8 h), and then they showed a similar EC relative to the base alloy using two-step aging (preaging at 90 °C + postaging 180 °C). However, the maximum allowable copper content is 0.1 wt.% for AA6201 and AA6101 conductor alloys according to the ASTM B398 and B317 standards [24, 25]. Therefore, the effect of the Cu addition beyond the specified value in the standards on EC could be controversial. On the other hand, the Ag addition (~ 0.5 wt.%) also led to a higher peak hardness than that of the base alloy in the Al-Mg-Si alloys [18-20], while the effect of Ag solutes on electrical resistivity was almost negligible [26]. To the best of the authors' knowledge, the effect of individual and combined Ag and Cu additions on the comprehensive properties in terms of strength and conductivity in Al-Mg-Si conductor alloys has been scarcely reported.

The present paper aims to investigate the efficiency of individual and combined additions of Ag and Cu in terms of the strength and electrical conductivity using conventional and modified thermomechanical treatments. The microstructural features were analyzed using transmission electron microscopy (TEM) and differential scanning calorimetry (DSC). To better understand the relationship between the strengthening factors and electrical resistivity sources, strength and electrical resistivity models were adopted in this study.

2. Experimental Procedure

The chemical compositions of the experimental alloys are listed in Table 1, in which the Mg, Si, and Fe levels were kept almost constant and the concentrations of Ag and Cu were varied. The alloy designation is based on the Ag and Cu levels. Commercial pure Al (99.85%), pure magnesium (99.8%), pure silver (99.9%), Al-50%Si, and Al-50%Cu master alloys (in wt.%) were used for the casting. After preparing the molten metal in an electrical resistance furnace, the melt was poured into a permanent steel mold with dimensions of 30 mm × 40 mm × 80 mm. Afterward, the as-cast alloys were homogenized at 560 °C for 6 h and then hot-rolled at 350-480 °C with an area reduction of 70%. Square bar samples (7 mm × 7 mm) were prepared from hot-rolled sheets for cold wire drawing. Two types of thermomechanical treatments were applied to the samples: conventional thermomechanical treatment (CTMT) and modified thermomechanical treatment (MTMT). A schematic representation of the thermomechanical treatments is shown in Fig. 1. CTMT was comprised of a solution heat treatment at 540 °C for 2 h, followed by quenching, cold wire drawing (70%), and post-aging at 180 °C for 0-24 h (Fig. 1a). MTMT was composed of a solution heat treatment at 540 °C for 2 h, quenching, pre-aging (6 h at 180 °C), cold wire drawing (70%), and post-aging at 180 °C for 0-24 h (Fig. 1b). It has been reported that 180 °C is a suitable postaging temperature to acquire a better trade-off between strength and EC in Al-Mg-Si conductor alloys [5]. In MTMT, pre-aging at 180 °C for 6 h is similar to the peak aging of the T6 temper. The cold wire drawing was performed with a die in six passes. The samples were water-cooled after each pass to avoid any temperature changes during the wire drawing. The final diameter of the wire was 4.37 mm after wire drawing.

Vickers microhardness was measured with a load of 25 g and a dwell time of 20 s. Eight indentations were made to obtain the average hardness values. The tensile strength was measured using an Instron tensile machine for wire samples with a 250 mm gauge length at strain rate of 1.6×10^{-3} s^{-1} according to the EN 50183 standard. At least three tensile specimens were tested to confirm the experimental results for each condition. The Sigmascope method with a frequency of 480 kHz was adopted to measure the electrical conductivity of the wires according to ASTM E1004.

Differential scanning calorimetry (DSC) was conducted at least three times for each condition under an argon atmosphere using a METTLER-TOLEDO system between 25 °C and 550 °C with a heating rate of 10 °C/min. The weight of DSC samples ranged from 15 mgr to 30 mgr. All DSC results were normalized by dividing the heat flow signal by the sample weight for each test. Disc-shaped specimens with a diameter of 3 mm for TEM were mechanically ground and then electropolished using a twin jet unit with an electrolyte of 1/3 HNO_3 in methanol at temperatures between -20 °C and -30 °C. The examination of the microstructures (precipitates and dislocations) was performed using TEM operated at 200 kV. The convergent beam electron diffraction (CBED) technique was then used to measure the specimen thickness. TEM micrographs were obtained with the Al matrix in the <001> zone axis for precipitates and in the <011> zone axis for dislocation observation. The methods for the quantitative analysis of the precipitates are described in detail in Ref. [8, 27].

Table 1 Chemical compositions of all studied alloys (wt.%).

Fig. 1 Schematic sketch for (a) conventional thermomechanical treatment (CTMT) (b) modified thermomechanical treatment (MTMT). Arrows show where the wire drawing was carried out.

3. Results

3.1. Evolution of mechanical strength and electrical conductivity

The hardness curves during isothermal artificial aging at 180 °C immediately after wire drawing are presented in Fig. 2a and 2c for CTMT and MTMT, respectively. As shown in Fig. 2a, all the curves displayed similar profiles for the alloys under CTMT: the hardness sharply increased and reached its maximum after 1 h or 2 h of aging, followed by an overaging stage with a decreasing hardness. The peak hardness of the CTMT base alloy was 108 HV. The Cu4 alloy exhibited the highest peak hardness (143 HV) among all alloys in CTMT (Fig. 2a). Apart from the Cu4 alloy, the Ag-/Cu-containing alloys exhibited a slightly higher peak hardness (3-7 HV) compared to the base alloy in CTMT: Ag3 (108 HV), Ag5 (112 HV), Cu1 (115 HV), Cu3 (115 HV), and Ag3Cu1 (115 HV). During the overaging stage, the hardness difference between the Ag-/Cu-added alloys and the base alloy became larger. In contrast to CTMT, the hardness of the MTMT alloys continuously decreased over the aging time because these alloys already reached the peak hardness before wire drawing (Fig. 2c). In general, the hardness values of the alloys processed with MTMT are higher than those of their counterparts processed with CTMT. Similar to CTMT, the Cu4 alloy showed the highest hardness values (156 HV) among all alloys using MTMT. The MTMT Ag3 (118-90 HV) and the MTMT base (117-88 HV) alloys showed a similar hardness behavior within 24 h (Fig. 2c). Concerning the other alloys, the aging hardening and the maximum hardness for the alloys containing Ag and Cu were greatly enhanced compared to the base alloy (5-20 HV): Ag5 (130 HV), Cu1 (122 HV), Cu3 (131 HV), and Ag3Cu1 (136 HV).

The hardness difference between peak aging and overging conditions for the base alloy was 22 HV, while the Ag-/Cu-added alloys showed (4-13) HV under CTMT (Fig. 2a). In addition, the hardness value in the MTMT base alloy decreased by 29 HV from peak aging to overging though that in the MTMT Ag-/Cu-added alloys reduced by (16-27) HV from peak aging to overging (Fig. 2c). It revealed that the Cu-added alloys exhibited a lower variation in the hardness in both CTMT and MTMT from peak aging to overging. It indicates a higher thermal stability in the Cu-added alloys, which is consistent with [21]. It might stem from the presence of the L phase with a superior thermal resistance in the Cu-added alloys [21].

Fig. 2b and 2d show the evolution of the electrical conductivity (EC) of the experimental alloys as a function of the aging times under CTMT and MTMT, respectively. All the investigated alloys showed a rapid increase in EC in the early stage of aging (up to 4 h), followed by a slight increase in EC (4-24 h) in both CTMT and MTMT (Fig. 2b and 2d). The order of the highest achievable EC of the CTMT alloys is the base (55.4% IACS), Ag3 (54.1% IACS), Ag5 and Cu1 (54% IACS), Cu3 and Ag3Cu1 (52.6% IACS), and Cu4 (49% IACS). Interestingly, the maximum EC of the MTMT alloys is 1-2% IACS higher than their CTMT counterparts: base (56.3% IACS), Ag3 (55.8% IACS), Cu1 (55.5% IACS), Ag5 (54.9% IACS), Cu3 and Ag3Cu1 (53.7% IACS), and Cu4

(50.9% IACS). Based on Fig. 2b and 2d, the EC levels under MTMT are higher than those under CTMT at a given aging time for all alloys, showing a more solute extraction in MTMT. In general, for both thermomechanical treatments, the higher the Cu content was, the more EC was limited. The adverse effect of Cu addition on EC could be observed from the Cu1, Cu3, and Cu4 alloys in Fig. 2b and 2d. Accordingly, the EC of the Cu4 alloy did not even reach the minimum required EC (52.5% IACS) in either CTMT or MTMT (Fig. 2b and 2d). This is most likely attributed to the high level of Cu solutes remaining in the matrix. The Cu3 and Ag3Cu1 alloys have the second lowest EC values in either CTMT or MTMT. It appears that Ag addition (in Ag3 and Ag5) showed a lower detrimental effect on EC than Cu addition.

By comparing the Cu4 alloy with other alloys (Fig. 2), it can be seen that there was a large hardness gap with the other alloys, either in CTMT or in MTMT. However, Cu4 alloys had the lowest EC for both thermomechanical treatments. In addition, the Ag3 and Cu1 alloys showed a similar combination of hardness and EC as the base alloy. This suggests that the effect of low Ag and Cu contents was not distinguishable on the hardness above 52.5% IACS. Therefore, the Ag5, Cu3, and Ag3Cu1 alloys were chosen for a further investigation because they showed a superior strength above 52.5% IACS compared to the base alloy.

Fig. 3 shows the microhardness versus EC above 52% IACS. Each individual aging time gives certain values of the hardness and electrical conductivity. Fig. 3 displays the plot of the hardness and electrical conductivity at all aging times. Although 52.5% IACS is the minimum required EC, 52% IACS was chosen as the starting point for the x-axis to have more span on EC for all alloys. The results indicate that the Cu-/Ag-added alloys (Ag5, Cu3, and Ag3Cu1) slightly outperformed (~5 HV) the base alloy under CTMT. Based on Fig. 3, the strength/EC windows of the CTMT Cu3 and Ag3Cu1 alloys are quite limited due to the lower EC (52.6% IACS), while the Ag5 alloy exhibits a broader window in the high end of strength and EC due to the relatively higher EC (54% IACS). The base, Ag5, Cu3, and Ag3Cu1 alloys in CTMT reached a minimum EC (52.5% IACS) after 2, 4, 24 and 24 h of aging, respectively. However, MTMT for the base alloy resulted in a higher hardness and electrical conductivity compared to the CTMT counterpart. Besides, the efficiency of the Ag and Cu additions was distinctly enhanced using MTMT, meaning that the hardness/EC plots of Cu-/Ag-added alloys lay considerably above that of the base alloy (~13 HV). Using MTMT, the Ag5, Cu3, and Ag3Cu1 alloys showed a similar hardness superiority above

52.5% IACS with overall higher values compared with the base alloy. The base, Ag5, Cu3, and Ag3Cu1 alloys in MTMT exceeded 52.5% IACS after 0.5, 2, 5, and 7 h of aging, respectively. On the other hand, the hardness/EC of all the MTMT alloys is superior to that of their CTMT counterparts, which is attributed to enhanced precipitation hardening [8]. In MTMT, the Cu3 and Ag3Cu1 alloys also had a narrower window of strength/EC relative to the Ag5 alloy, but they all showed a higher strength/EC compared to their counterparts in CTMT.

Fig. 2 Evolution of (a) HV hardness and (b) electrical conductivity (EC) as a function of the aging time using CTMT; (c) HV hardness and (d) electrical conductivity (EC) as a function of the aging time using MTMT.

Fig. 4a shows the maximum strength values of the base, Ag5, Cu3, and Ag3Cu1 alloys under CTMT and MTMT above 52.5% IACS. Using CTMT, the ultimate tensile strengths for the base, Ag5, Cu3, and Ag3Cu1 alloys are 296, 306, 316, and 305 MPa, respectively, showing that the strength were slightly increased (~10-20 MPa) with Ag/Cu additions above 52.5% IACS compared to the base alloy (Fig. 4a). However using MTMT, the ultimate tensile strengths for the base, Ag5, Cu3, and Ag3Cu1 alloys are 317, 341, 348, 343 MPa, respectively, indicating that MTMT led to considerably higher strength levels (~30-35 MPa) with Ag/Cu additions relative to the base alloy, which is consistent with the hardness results. It is worth mentioning that the MTMT Ag5, MTMT Cu3, and MTMT Ag3Cu1 alloys exhibited a remarkably higher strength by ~50 MPa (18%) compared to the CTMT base alloy. Fig. 4b displays a comparison of the strength/EC of the investigated alloys with values in the EN50183 standard (AL2-AL6). It is apparent that the MTMT Ag5, MTMT Cu3, and MTMT Ag3Cu1 alloys show remarkably higher strengths than the values in the EN50183 standard above 52.5% IACS. In contrast, Cu-/Ag-added alloys under CTMT exhibited similar strength/EC levels relative to those in the EN50183 standard. This suggests that MTMT led to maximizing the efficiency of Cu/Ag additions in terms of the strength/EC.

Fig. 3 HV hardness as a function of EC above 52% IACS for all alloys using conventional and modified thermomechanical treatments.

Fig. 4 (a) the maximum ultimate tensile strengths above 52.5% IACS for the base, Ag5, Cu3, and Ag3Cu1 alloys treated by conventional and modified thermomechanical treatments; (b) a comparison between the strengths of the studied alloys and the values defined in EN50183 (AL2-AL6) standard for the wire with a diameter of 4.37mm

3.2. DCS analysis

Thermal events associated with precipitation behavior were studied via a DSC analysis. Fig. 5a and 5b show the DSC curves for the base, Cu3, Ag5 and Ag3Cu1 alloys in the as-quenched and as-drawn conditions, respectively. The DSC analysis in the as-quenched condition corresponds to the precipitation sequences in the preaging (MTMT), while the results in the as-drawn state uncovered the precipitation events in CTMT. In Fig. 5a, exothermic peaks I, II, and III could be correlated with the precipitate formation of β'', β', and β, respectively, which is consistent with the DSC results in [8, 28]. A closer look at exothermic peak I revealed that the peak temperatures of the β'' precipitation for the base, Ag5, Cu3, and Ag3Cu1 alloys are 273, 264, 270, and 254 °C, respectively. This implies that the Ag/Cu additions shifted the β'' peak to lower temperatures, indicating an accelerated precipitation rate. The Ag addition resulted in a higher precipitation rate than the Cu addition, as confirmed in [20]. The higher precipitation kinetics in the Ag5 alloys could correspond to a higher diffusion coefficient of Ag in Al relative to all other alloying elements. In addition, Ag has a strong interaction energy with Mg and vacancies, which is favorable for precipitation. Copper also has a strong interaction energy with Mg, but it has a relatively lower diffusion coefficient in aluminum [18]. It was reported that the Cu/Ag addition led to the formation of stable clusters with higher densities and larger sizes in the early stage of aging, evolving into copious subsequent precipitates [18, 19]. Based on Fig. 5a, no dissolution troughs were found in the Cu3 and Ag3Cu1 alloys before the β" precipitation peak, showing more stable β" precursors (GP zones). Comparing the as-quenched base alloy with the as-quenched Ag5 alloy (Fig. 5c), it can be seen that the dissolution trough (IV) in the Ag5 alloy is smaller than that in the base sample, suggesting more stable GP zones in the Ag5 alloy. In addition, the Ag5 alloy exhibited a small precipitation peak (V) at ~217 °C between the dissolution trough (IV) and β" precipitation peak (I), which could correspond to the formation of the pre-β" phases. Therefore, it might cause a slightly smaller β" peak in the Ag5 alloy compared to the base alloy, which might be related to a partial consumption of the precipitation capability in precipitation peak V [20]. It is worth mentioning that precipitation peak II became larger with the addition of Ag and Cu (Fig. 5a), which might be ascribed to the formation of Ag-/Cu-containing precipitates. Similar results were acquired in [20, 21].

Fig. 5b displays the DSC curves obtained in the as-drawn condition. As discussed in our previous works [8, 10], heavy cold deformation after solution treatment/quenching might disrupt the precipitation event. Accordingly, β' precipitation was promoted along the dislocations in the drawn samples at the expense of the β" phases, as reported in Ref. [29-31]. Therefore, peak I_d could correspond to the simultaneous formation of needle-shaped and lath-like precipitates with a disordered β' structure. The exothermic II_d peak was associated with the formation of the β phase [29-31]. It should be noted that materials with a high area reduction (70%) in wire drawing exhibited exothermic recrystallization peaks (R_x) at approximately 400 °C, which overlapped with the β precipitation peak, which is consistent with Ref. [30]. The peak temperatures of β' precipitation (i.e., peak I_d) for the base, Ag5, Cu3, and Ag3Cu1 alloys are 236, 236, 233, and 235 °C, respectively, showing a lower impact of the Ag/Cu addition on the precipitation rate. In addition, the Ag-/Cu-added alloys showed a slightly larger area of the β' peak (I_d) compared to the base alloy. Fig. 5b reveals that no dissolution troughs and no precipitation peaks were observed before peak I_d either in the Ag-/Cu-added alloys or in the base alloy. It can be deduced that the presence of the dense network of dislocations could somehow counteract the effect of Ag and Cu additions either in precipitation kinetics or in precipitation hardening, which is consistent with a lower increased peak hardness (3-7 HV) with Ag and Cu additions in CTMT. It should be noted that the Cu3 alloy revealed an extra peak (III_d) in the as-drawn condition at 260-290 °C (Fig. 5b). This might correspond to the formation of the Q' phase, which more easily forms in local defective zones, such as dislocations [29, 32].

A comparison of exothermic peaks I (as-quenched condition) and I_d (as-drawn condition) for the base alloy is shown in Fig. 5c. It can be deduced that heavy cold deformation in wire drawing could lead to a smaller first precipitation peak, meaning that wire drawing could weaken the precipitation hardening response. However, precipitation kinetics significantly increased when applying the wire drawing.

Fig. 5 DSC curves of the base, Cu3, Ag5 and Ag3Cu1 alloys in (a) as-quenched condition and in (b) as-drawn condition; (c) an enlarged view for the as-quenched base alloy, as-quenched Ag5 alloy and as-drawn base alloy; AQ stands for “as-quenched” and AD stands for “as-drawn”.

3.3. Precipitate observation by TEM

Fig. 6 shows the TEM bright-field micrographs of the four alloys (the base alloy after 2 h aging, Ag5 alloy after 4 h aging, Cu3 alloy after 24 h aging and Ag3Cu1 alloy after 24 h aging) using CTMT. It should be noted that these sample conditions are all above 52.5% IACS. All the TEM images were acquired in the <002>Al directions, showing that all precipitates extended along the <200>Al directions of the matrix. As discussed in our previous work [8], the dominant microstructural feature for CTMT alloys is large lath-like precipitates and needle-like precipitates with a disordered β' structure. In the Cu3 alloy, it is speculated that there is a small number of Q' phases, which corresponds to the extra peak (III_d) in Fig. 5b. The precipitate statistics (density, length, and cross section) under these conditions were calculated and are shown in Fig. 6e and 6f. From Fig. 6e, it is fairly evident that the Ag/Cu additions led to an increase in the precipitate number density by 3-4 × 10^3 μm^{-3} in the CTMT Ag5, Cu3, and Ag3Cu1 alloys compared to the CTMT base alloy. The increased precipitate number density could correspond to the formation of Ag-/Cu-containing precipitates [32]. Therefore, the moderately higher strength in the Ag5, Cu3, and Ag3Cu1 alloys compared to the base alloy under CTMT (Figs. 3 and 4) could be ascribed to the relatively higher number density of precipitates. All alloys showed similar cross-sectional areas of the precipitates, but the average precipitate length in the Ag-/Cu-added alloys was relatively finer than that in the base alloy, which was also reported in Ref. [19]. It should be noted that the identification of the exact nature of the precipitates, which might be influenced by Ag and Cu additions, is beyond the scope of this study.

Fig. 6 Bright-field TEM micrographs of (a) base alloy after 2 h aging (b) Ag5 alloy after 4 h aging (c) Cu3 alloy after 24 h aging (d) Ag3Cu1 alloy after 24 h aging using the conventional thermomechanical treatment; (e) the number density of precipitates (f) the average length and the cross section of precipitates. C stands for conventional thermomechanical treatment.

Fig. 7 shows TEM micrographs of the base, Ag5, Cu3, and Ag3Cu1 alloys immediately after preaging (6 h at 180 °C). With the addition of Ag and Cu, the average number density of the precipitates increased, and the average length of the precipitates decreased in the Ag5, Cu3, and Ag3Cu1 alloys compared to the base alloy (Fig. 7e and 7f). Similar results were reported in Ref. [19], indicating that Ag and Cu promote precipitation. The predominant precipitates in the preaged base alloy were β'' precipitates due to their morphology and sizes [33]. It was reported that the addition of 0.5 wt.% Ag to Al-Mg-Si alloys did not change the type of precipitates, but it did lead to a higher number density of precipitates with a shorter size [20], as shown in Fig. 7. Kim et al. [34] reported that β'' precipitates were the predominant precipitates in Al-Mg-Si alloys containing 0.21 wt. %Cu. Therefore, it is deduced that the β'' phases are the dominant precipitates in the Cu-added alloys. Accordingly, the higher hardness values of Ag5 (117 HV), Cu3 (118 HV), and Ag3Cu1 (121 HV) alloys compared to the base alloy (107 HV) after preaging could be supported by their precipitate number density (Fig. 7e).

Fig. 8 displays the TEM images of the four alloys (base alloy after 0.5 h of aging, Ag5 alloy after 2 h of aging, Cu3 alloy after 5 h of aging, Ag3Cu1 alloy after 7 h of aging) under MTMT. It is obvious that the Ag and Cu additions in the Ag5, Cu3, and Ag3Cu1 alloys resulted in a remarkably increased precipitate number density by 7-9 × 10^3 μm^{-3} compared to the base alloy using MTMT (Fig. 8e). All MTMT alloys exhibited a similar precipitate length and cross-sectional area (Fig. 8f). Therefore, the maximized efficiency of Ag and Cu under MTMT in terms of the strength above

52.5% IACS could be ascribed to the outstandingly increased number density of precipitates. In MTMT, the preaged precipitates could be changed during further wire drawing and postaging. The TEM results (Fig. 8) reveal that the precipitate sizes were more or less similar to the preaged condition for all the MTMT alloys after wire drawing. This result is likely attributed to the shearing effect on the precipitates during cold wire drawing being compensated by the coarsening of the precipitates during postaging in the overaged condition. In general, the MTMT alloys showed a denser distribution of remarkably short precipitates than the CTMT alloys, suggesting promising precipitation strengthening in the MTMT alloys. Comparing the number density of precipitates under two thermomechanical treatments (Fig. 6e in CTMT and Fig. 8e in MTMT), it is apparent that the number density of the precipitates in the MTMT alloys is much higher by more than three times than that in the CTMT alloys. It could also be inferred that wire drawing in CTMT deteriorated the precipitation hardening. This was also confirmed with the DSC results (Fig. 5c); the first precipitation peak (I_d) in the as-drawn base sample (CTMT) was smaller than the first precipitation peak (I) in the as-quenched base sample (MTMT).

Fig. 7 Bright-field TEM micrographs of (a) base alloy (b) Ag5 alloy (c) Cu3 alloy and (d) Ag3Cu1 alloy, after 6 h pre-aging without wire drawing, (e) the number density of precipitates (d) the average length and the cross section of precipitates. BD stands for before drawing.

Fig. 8 Bright-field TEM micrographs of (a) base alloy after 0.5 h aging (b) Ag5 alloy after 2 h aging (c) Cu3 alloy after 5 h aging (d) Ag3Cu1 alloy after 7 h aging using the modified thermomechanical treatment; (e) the number density of precipitates (f) the average length and the cross section of precipitates. M stands for modified thermomechanical treatment.

Fig. 9 shows the distribution of the partially deviated precipitates in the <001> zone axis for Ag3Cu1 after 7 h of aging and the Cu3 alloy after 5 h of aging under MTMT. In the enlarged areas

(Fig. 9a and 9b), the deformed bands of the partially disordered precipitates could be observed in both the Cu3-5h and Ag3Cu1-7h alloys. As shown in Fig. 6f and 8f, shorter precipitates were acquired in MTMT than in CTMT. This is consistent with our previous findings [8], showing that the pre-existing precipitates formed by pre-aging were fragmented during wire drawing. Fig. 9 shows that the precipitates were repeatedly sheared by the moving dislocations due to the high degree of cold deformation, causing the partially deviated precipitate band. The same observations were also observed in the base-0.5h and Ag5-2h alloys.

Fig. 7 TEM micrographs to show the partially deviated precipitates in <001> zone axis in (a) MTMT Ag3Cu1 after 7 h aging and (b) MTMT Cu3 after 5 h aging.

3.4. Effect of wire drawing on the strength and EC

Strain hardening is one of the most promising strengthening mechanisms in Al-Mg-Si cables, affecting the strength and electrical conductivity. Fig. 10 shows the increased hardness and the decreased EC due to wire drawing. The increased hardness and the reduced EC could be obtained by measuring the hardness and EC difference immediately before and immediately after wire drawing without postaging treatment. The increased hardness for the CTMT base alloy and the MTMT base, MTMT Cu3, MTMT Ag5 and MTMT Ag3Cu1 alloys were 26, 10, 13, 13, and 15 HV, respectively. It is assumed that the softening recovery would be low during aging at 180 °C, as aging at 180 °C for up to 34 h showed negligible recovery softening on commercially pure drawn aluminum [10]. It is deduced that the strain hardening contribution of the base alloy under CTMT is almost double relative to the alloys subjected to MTMT, which was also confirmed in our previous work [8] and by Ref. [35, 36]. The dislocation networks of the alloys are shown in Fig. 11 to correlate the work hardening contributions with the microstructure. In general, a dislocation structure with a highly complex and entangled network could be created due to the large plastic deformation. As shown in Fig. 11, a denser network of dislocations was created in the

CTMT base alloy during wire drawing compared to all the MTMT alloys, which is consistent with their strain hardening contributions.

Fig. 8 The increased hardness and decreased electrical conductivity by wire drawing in the base alloy under CTMT and MTMT conditions and the MTMT Cu3, Ag5, and Ag3Cu1 alloys.

As displayed in Fig. 10, the reduced EC via wire drawing in the CTMT base, MTMT base, MTMT Cu3, MTMT Ag5 and MTMT Ag3Cu1 alloys was 0.25, 0.65, 1.3, 1.2, and 1.0% IACS, respectively. It is evident that the reduced EC in the alloys under MTMT is higher than that in the alloys under CTMT. According to the increased hardness values, the dislocation number densities created in the MTMT alloys are less than those in the CTMT alloy, contradicting the reduced EC values. The possible reason for the higher reduced EC in the MTMT alloys could be related to a slight redissolution of β'' precipitates during cold deformation. In our previous work [8], it was determined that cold wire drawing led to a slight decrease in the precipitate number density, which corresponded to the redissolution of the β'' phases during drawing. The redissolution of the precipitates during cold deformation was also reported in Ref. [36, 37]. In addition, it was reported that the denser dislocation zone showed a reduced precipitate density in regions after cold deformation of pre-aged AA6101 [37], which might be an indication of the redissolution of β'' precipitates.

Fig. 9 The dislocation cell substructures for (a) CTMT base after 2 h aging; (b) MTMT Cu3 after 5 h aging; (c) MTMT Ag5 after 2 h aging; (d) MTMT Ag3Cu1 after 7 h aging. All were taken from <011> zone axis close to (111) planes.

4. Discussion

4.1. Strength and electrical resistivity models

To elucidate the dominant strengthening mechanism for Al-Mg-Si conductor alloys, a theoretical analysis was quantitatively conducted based on the microstructural features. The principal strengthening factors in Al-Mg-Si alloys come from the grain boundaries, dislocations, precipitates, and solid solution hardening. Given that the yield strength values (in MPa) are roughly equal to three times the hardness (in HV), the contribution of these mechanisms can be linearly added to the total hardness through the following equation [38, 39]:

$$HV_{total} = HV^{Al} + \Delta HV^{dislo} + \Delta HV^{prec} + \Delta HV^{gb} + \Sigma\Delta HV_i^{sol} \quad (1)$$

Where HV_{total} is the total hardness, HV^{Al} is the aluminum matrix (12 HV, considering the yield stress of an annealed pure 1100-O aluminum alloy is ~35 MPa for [40]) and ΔHV^{dislo}, ΔHV^{prec}, ΔHV^{gb} and $\Sigma\Delta HV_i^{sol}$ are the hardness contributions caused by dislocations, precipitates, grain boundaries, and solid solutions, respectively [38, 39, 41].

The main strain hardening portion was created during wire drawing. As discussed in Section 3.4, the strain hardening hardness values for the CTMT base alloy were 26 HV, and those for the MTMT base, Cu3, Ag5 and Ag3Cu1 alloys were 10, 13, 13, and 15 HV, respectively. Using the Bailey-Hirsch relationship (Eq. 2), the dislocation number density could be estimated for each alloy [41, 42]. Accordingly, the approximate dislocation number densities for the CTMT base alloy and for the MTMT base, Cu3, Ag5 and Ag3Cu1 alloys are 2.8×10^{14} m^{-2}, 0.4×10^{14} m^{-2}, 0.7×10^{14} m^{-2}, 0.7×10^{14} m^{-2} and 0.9×10^{14} m^{-2}, respectively. Later, the dislocation contributions to the electrical resistivity could be calculated for each alloy based on the dislocation number densities.

$$\Delta\sigma^{dislo} = \alpha.M.G.b.\rho^{1/2} \quad (2)$$

Where ρ is the dislocation number density [42], M = 2 is the Taylor factor [43], α is a constant (0.3 for Al), G is the shear modulus (26.9 GPa for Al), and b is the Burgers vector (2.9 A° for Al).

The approximation of the precipitation hardening could be calculated through the Orowan bypass mechanism. For the CTMT base alloy, the predominant disordered β' precipitates were characterized as non-shearable precipitates in the overaged condition (Fig. 6), as discussed in Ref. [8]. For MTMT alloys, identifying the precipitate's nature might be difficult based on their morphology and size due to the shearing of precipitates during cold wire drawing and the coarsening of precipitates during post-aging [8]. Since all MTMT alloys were in the overaged condition when they reached the minimum required EC, it is reasonable to assume that the dominant precipitates were non-shearable in MTMT. Therefore, the Orowan equation (Eq. 3) could be adapted to calculate the precipitation hardening.

$$\Delta\sigma_{Or}{}^{prec} = \frac{2M\beta Gb}{L} \tag{3}$$

Where M, G and b were defined in Eq. 2, β is a dimensionless constant (0.28), and L is the interprecipitate spacing, which is measured as $L= \frac{1}{(N)^{1/3}}$ (where N is the precipitate number density) [42]. For the contribution of precipitates to the hardness of the CTMT base alloy and the MTMT base, the Cu3, Ag5 and Ag3Cu1 alloys are estimated to be 52.3, 82.9, 90.7, 92.2, and 90.2 HV, respectively.

The hardness increment from the grain boundary strengthening could be approximated by the classical Hall-Petch equation (Eq. 4) [42].

$$\Delta HV^{gb} = k_{HP}d^{-1/2} \tag{4}$$

In this relationship, k_{HP} is the Hall-Petch coefficient (35 HV $\mu m^{1/2}$) for 6xxx series aluminum alloys, and d is the average grain size [42]. Since the average grain sizes were measured as ~380 µm for all alloys, the contribution of the grain boundary strengthening is estimated as ~2 HV for all alloys, which is negligible compared to the total hardness.

Solid-solution strengthening can be calculated using the following equation [44].

$$\Delta\sigma_i{}^{sol} = k_iC_i{}^{2/3} \tag{5}$$

In this equation, k_i is the solid solution strengthening coefficient ($k_{Mg} \approx 29$, $k_{Si} \approx 66.3$, and $k_{Cu} \approx 46.4$ MPa (wt.%)$^{-2/3}$), and C_i is the concentration of solutes in solid solution form [44]. Given that the atom sizes of Ag (144 A°) and Al (143 A°) are close [18, 19, 26], it seems that a solid solution of Ag in Al alloys has a negligible strengthening effect. According to our previous work [10], it was assumed that half of the Mg might remain in the matrix of the alloy with Mg/Si ~ 2. In addition, the Si level was considered to be negligible in Mg-excessive alloys (Mg/Si ~ 2). Li et al. [45] reported that ~ 20% of the Cu addition could be taken apart in the precipitation and that ~ 80% of the Cu atoms might be left in the matrix. Accordingly, the solute strengthening for the CTMT base alloy and the MTMT base, Cu3, Ag5 and Ag3Cu1 alloys could be approximately calculated as 5, 5, 10, 5, and 8 HV, respectively.

A summary of the predicted contribution of different hardening mechanisms and a comparison with the experimentally measured hardness values is shown in Fig. 12. The predicted total hardness values were generally in good agreement with the experimentally measured values under various conditions. The slightly higher predicted hardness in the MTMT alloys (Ag5, Cu3 and Ag3Cu1) might be correlated to the overestimated solute strengthening. Precipitation and strain hardening play prominent roles in strengthening with an ~80% contribution of the total hardness. The MTMT led to the enhanced precipitation hardening at the expense of strain hardening relative to the CTMT base alloy. As shown in Fig. 12, with the additions of Ag and Cu (Cu3, Ag5, Ag3Cu1 alloys), the precipitation hardening was further improved under MTMT. The increased precipitation hardening in the MTMT alloys is superior to the reduced strain hardening. The precipitation strengthening accounts for ~80% of all strengthening mechanism contributions, except for the aluminum matrix.

Fig. 10 The comparison between predicted and experimentally measured hardness for CTMT base alloy after 2 h aging, and for the MTMT base after 0.5 h aging, Ag5 after 2 h aging, Cu3 after 5 h aging and Ag3Cu1 after 7 h aging. C and M stand for the conventional and modified thermomechanical treatment, respectively.

Similar to the strengthening model, it is worth quantifying the electrical resistivity coming from the various microstructural features using Matthiessen's rule (Eq. 6) [41, 42, 46].

$$\rho_{total} = \rho^{Al\ pure} + \rho^{DISLO} + \rho^{PREC} + \rho^{GB} + \rho^{SOL} \qquad (6)$$

Where ρ_{total} is the total electrical resistivity, $\rho^{Al\ pure}$ is the electrical resistivity of commercially pure aluminum (AA1350–O) (2.79×10^{-6} Ω cm at room temperature) [24], and ρ^{DISLO}, ρ^{GB}, ρ^{PREC} and ρ^{SOL} are the electrical resistivity contributions from dislocations, grain boundaries, precipitates and solute atoms, respectively [41, 42, 46]. Eq. 7 could be employed to convert the electrical resistivity (μΩ cm) to electrical conductivity (% IACS) [47].

$$\text{EC (\% IACS)} = \frac{172.4}{\rho tot \quad (\mu\Omega\ \text{cm})} \tag{7}$$

The contribution of dislocations to the electrical resistivity could be estimated as [42].

$$\Delta\rho^{DISL} = L^{dislo}\Delta\rho^{dislo} \tag{8}$$

Where L^{dislo} is the dislocation density in m^{-2}, $\Delta\rho^{dislo}$ is the electrical resistivity per unit density of dislocations ($2.7\times10^{-25}\ \Omega m^3$) [42]. Given the dislocation densities for the CTMT base alloy, and for the MTMT base, MTMT Cu3, MTMT Ag5 and MTMT Ag3Cu1 alloys, the electrical resistivities are 0.008, 0.001, 0.002, 0.002, and 0.003 μΩ cm respectively, which is negligible compared to the total electrical resistivity.

The effect of precipitates on electrical resistivity of Al can be calculated through Eq. 9 [48].

$$\Delta\rho^{PREC} = \frac{\Delta\rho^{prec}}{(L^{prec})^{1/2}} \tag{9}$$

In this equation, $\Delta\rho^{prec}$ is the precipitate resistivity constant ($12\ \Omega(nm)^{3/2}$), and L^{Prec} is the average interprecipitate spacing (in nm) [48]. Considering the space between the precipitates, the contributions of precipitates to the electrical resistivity for the CTMT base alloy and for the MTMT base, the Cu3, Ag5 and Ag3Cu1 alloys were calculated to be 0.16, 0.20, 0.21, 0.21, and 0.21 μΩcm, respectively.

The impact of grain boundaries on the electrical resistivity can be evaluated as follows [42].

$$\Delta\rho^{GB} = S^{GB}\Delta\rho^{gb} \tag{10}$$

In this relationship, ($S^{GB} = 6/d$) is the bulk density of the grain boundaries with an average grain size (d), and $\Delta\rho^{gb}$ is the resistivity per unit concentration of grain boundaries ($2.6\times10^{-16}\ \Omega m^2$) [42]. Considering the similar grain size for all alloys (~380 μm), the contribution of grain boundaries to resistivity is calculated as 0.0004 μΩ cm, which is also negligible compared to the total electrical resistivity.

The contribution of the solid solution to the electrical resistivity can be calculated by Eq. 11 [42].

$$\Delta\rho^{SOL} = \Sigma C_i^{sol}\Delta\rho_i^{sol} \tag{11}$$

Where C_i^{sol} is the solute concentration, and $\Delta\rho_i^{sol}$ is the resistivity per unit contribution of solute i, which is 0.445, 0.496, and 0.299 μΩ cm (at. %)$^{-1}$ for Mg, Si [42], and Cu [49], respectively.

Considering the similar atom radii of Ag and Al, a negligible electrical resistivity of the Ag solute in the Al matrix is assumed [26]. Based on the solute levels of Mg, Si and Cu, the solute electrical resistivities for the CTMT base alloy and for the MTMT base, Cu3, Ag5 and Ag3Cu1 alloys were estimated to be 0.14, 0.14, 0.21, 0.14, and 0.17 μΩ cm, respectively.

The total estimated electrical resistivities for the CTMT base alloy, and for the MTMT base, Cu3, Ag5 and Ag3Cu1 alloys are 3.101, 3.136, 3.212, 3.148, and 3.180 μΩ cm, respectively. Fig. 13 shows the contribution of the different factors to the electrical resistivity. It is apparent that the total electrical resistivity is mainly controlled by the solutes and precipitates for all alloys. The contributions of grain boundaries and dislocations to electrical resistivity are too small to become visible. The electrical resistivities of precipitates for the MTMT alloys are higher than the CTMT base alloy due to their higher precipitate number densities. Comparing Figs 12 and 13, it is found that the dislocations (stain hardening) showed a considerable contribution to the hardness but with a negligible effect on the electrical resistivity. As seen in Fig. 12, the precipitate hardening contribution in the CTMT base is 52.3 HV, and those in the MTMT Cu3, Ag5 and Ag3Cu1 alloys are 90.7, 92.2, and 90.2 HV, respectively. It shows that the precipitate strengthening was enhanced by ~75% by applying MTMT and adding Ag and Cu. On the other hand, the precipitate resistivity portion in the CTMT base alloy is 0.16 μΩcm while those in the MTMT Cu3, Ag5 and Ag3Cu1 alloys are 0.21 μΩcm, suggesting that applying MTMT and adding Ag and Cu led to an increase in electrical resistivity by ~30% (Fig. 13). Therefore, precipitate strengthening is the key factor to increase strength with less detrimental effect on the electrical conductivity.

Fig. 14 displays the plot of the hardness/EC for the predicted and experimental values for all alloys. The results of the hardness/EC predicted by the strength/electrical resistance models reflect the general tendency of the hardness/EC changes influenced by the main strengthening mechanisms, but the predicted values shift toward higher EC values relative to the experimental values. The reason is mainly related to the electrical resistance model, in which the predicted EC was at most 5% higher than the experimentally measured values. Based on Eq. 7, the estimated EC values for the CTMT base alloy and for the MTMT base, Cu3, Ag5 and Ag3Cu1 alloys are 55.59, 54.97, 53.66, 54.76, and 54.21% IACS, respectively, while the experimental EC values are 53, 53.3, 52.7, 53, and 52.7% IACS, respectively. It is worth noting that the electrical resistance of aluminum alloys is very sensitive to any crystal and microstructure change. Some crystal and microstructure parameters required in the electrical resistance model are difficult to determine and rely on the best estimation, which may vary over a relatively wide range. For instance, it is difficult to measure the solid solution level for each alloying element in a multicomponent aluminum system, even using advanced experimental techniques. However, any change in the solid solution level may result in a large change in the electrical resistance. Apparently, the current electrical resistance model based on Matthiessen's rule is still semiquantitative and needs further improvement.

Fig. 11 Electrical resistivity contributions of the different strengthening factors for all alloy conditions. The resistivities of grain boundaries and dislocations are almost not visible due to their small portions. C and M stand for the conventional and modified thermomechanical treatment, respectively.

Fig. 12 The comparison of the hardness/EC values between the predicted and experimentally measured ones. C and M stand for the conventional and modified thermomechanical treatment, respectively.

4.2. Correlation between strength and electrical conductivity

As shown in Fig. 2, the Ag and Cu additions showed a remarkable efficiency in terms of the peak hardness in MTMT (12-20 HV) and preaged conditions (10-14 HV) relative to CTMT (3-7 HV). Comparing the peak temperatures of the first precipitation peak in the as-drawn condition (i.e., peak I_d) with the as-quenched state (i.e., peak I), the copious dislocations in the CTMT alloys could somehow counteract the effect of the Ag and Cu additions on precipitation kinetics. The slightly increased peak hardness (3-7 HV) in the Cu3, Ag5 and Ag3Cu1 alloys relative to the base alloy using CTMT shows the dominant impact of the dislocations on the precipitation rather than Ag and Cu additions. Moreover, the DSC results revealed that the large quantities of dislocations prior to aging (the CTMT base alloy) could weaken the precipitation hardening response compared to the MTMT counterpart (Fig. 5c), which was confirmed by the TEM analysis in Fig. 6-8.

The maximum EC levels in the MTMT alloys are higher than those in the CTMT counterparts by (1-2)% IACS, suggesting that the modified thermomechanical treatment led to more purification of the matrix and a higher EC for all alloys. It should be noted that the atom radii of Al, Ag, and Cu are 1.43, 1.44, and 1.28 A°, respectively; hence, the zone misfits of Ag and Cu in the aluminum matrix are +0.7% and -10.5%, respectively [50]. Therefore, Cu could have a potentially more detrimental effect on EC than Ag, as discussed in Section 3.1. Accordingly, the alloy containing 0.42 wt.% Cu (Cu4) did not reach the minimum EC (52.5% IACS) under CTMT after 24 h of aging, while the alloy containing 0.53 wt.% Ag (Ag5) exceeded 52.5% IACS using CTMT after 4 h of aging. These results confirm that a high level of Cu in the Al-Mg-Si conductor alloy significantly limited EC in CTMT. However, MTMT could overcome the restriction of EC in Cu-added alloys, to some extent. The MTMT alloys met 52.5% IACS at a shorter postaging time relative to the CTMT alloys (particularly Cu-added alloys), which could be an economical benefit. For instance, the Cu3 and Ag3Cu1 alloys met the minimum required EC in CTMT after 24 h of aging but only after 5 and 7 h of aging in MTMT, respectively.

As displayed in Fig. 3, the MTMT base alloy showed a higher strength/EC than the CTMT base alloy. Furthermore, using MTMT, the Cu/Ag additions (the Ag5, Cu3 and Ag3Cu1 alloys) led to a further increased strength above 52.5% IACS relative to the base alloy. Therefore, the MTMT alloys generally exhibited a higher strength/EC than their counterparts under CTMT due to the enhanced precipitation hardening. In the traditional aging treatment (solution, quenching and aging) or preaging in this study, the contribution of the Ag/Cu atoms to the precipitation was observed with the addition of Ag and Cu [18, 19]. It was reported that the Ag and Cu atoms could remove some of the Mg atoms to form Ag/Cu-Mg clusters due to the strong interaction between Ag/Cu atoms and Mg atoms, evolving into Mg-Si-Ag/Cu co-clusters. These clusters along with the Mg-Si clusters resulted in a higher number density of stable precursors for the subsequent precipitates in Ag-/Cu-added alloys [18, 19]. It was expected that preaging in MTMT led to more Ag/Cu atoms being incorporated into copious precipitates to a significant extent, as confirmed in Ref. [18, 19]. Therefore, it is deduced that MTMT resulted in more extraction of the solutes from the aluminum matrix into the numerous precipitates, leading to a concurrently higher strength and higher EC levels.

In contrast, dense dislocations are the main favorable nucleation sites for precipitates in CTMT, particularly lath-like precipitates [8]. As mentioned earlier, Ag/Cu addition led to a slight increase in the peak hardness (Fig. 2) and hence a relatively higher precipitate number density under CTMT (Fig. 6). Therefore, it is suggested that the presence of dense dislocations might partly neutralize the efficiency of Ag/Cu atoms in clustering and precipitation.

5. Conclusions

1. The Ag and Cu additions had a relatively low impact on the strengths above the minimum required electrical conductivity (52.5% IACS) using a conventional thermomechanical treatment (CTMT) due to the strong effect of dislocations on the precipitation. However, the modified thermomechanical treatment (MTMT) maximized the strengthening effect of Ag and Cu additions on precipitation above 52.5% IACS. In addition, the postaging time was considerably reduced in order to reach the minimum required electrical conductivity (52.5% IACS) in MTMT compared to CTMT.

2. An increase in the Cu levels remarkably limited electrical conductivity in both CTMT and MTMT (particularly above 0.3 wt.% Cu) via the scattering electron by the remaining Cu solutes in the aluminum matrix. However, a 0.5 wt.% Ag addition in the Al-Mg-Si alloys still exhibited a reasonable electrical conductivity due to its similar atom radius to Al atoms and the small zone misfit of Ag in the aluminum matrix.

3. MTMT resulted in a wider window in the high end of strength and electrical conductivity compared to CTMT for all alloys. Pre-aging in MTMT maximized precipitation hardening at the expense of strain hardening. The superiority of precipitation hardening in the MTMT alloys led to a higher strength compared to the CTMT counterparts above the minimum electrical conductivity.

4. An improvement in precipitation strengthening could be a promising avenue in MTMT for obtaining a strength above the minimum electrical conductivity for an optimal compromise between the conflicting mechanical and the electrical properties of aluminum conductors. For

instance, Ag and Cu addition under MTMT led to an increase in the precipitate hardening contribution by 75%, while the resistivity portion of the precipitates only increased by 30%.

5. Strength and electrical resistance models were used to account for the multiple contributions of the main strengthening mechanisms and microstructure features to the mechanical strength and electrical conductivity, providing a useful tool for designing aluminum conductors and optimizing the mechanical and electrical properties.

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CRediT authorship contribution statement

Siamak Nikzad Khangholi: Methodology, Investigation, Formal analysis, Writing – original draft. **Mousa Javidani:** Conceptualization, Methodology, Writing – review & editing. **Alexandre Maltais:** Conceptualization, Validation, Writing – review & editing. **X.-Grant Chen:** Conceptualization, Validation, Writing – review & editing, Project administration.

Declaration of competing interest

The authors declare that they have no known competing financial interests or personal relationships that could have appeared to influence the work reported in this paper.

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