

Corporation, were fabricated as 130 mm diameter billets by blending helium inert gas atomized pre-alloyed powder with F-8 grade SiC whiskers, cold compacting and vacuum hot pressing in the mushy, liquid plus solid phase field. Following compaction, the alloys were hot extruded to 12.7 mm thick by 127 mm wide planks, cut to 200 mm length and hot cross rolled to 2 mm sheet.

Natural and artificial aging studies involved solution treatment for 30 min. in an argon atmosphere at $525^{\circ}\text{C} \pm 3^{\circ}\text{C}$ followed by cold water quenching and immediate aging. The kinetics of aging were evaluated by measurement of the electrical conductivity (percent IACS) and the Rockwell B hardness, the latter being reported as the average of five readings. Based on these results selected conditions were prepared for transmission electron microscopy utilizing a JEOL 100C. These samples were initially ground to 0.1 mm thickness with final preparation by argon ion thinning.

Table I. Chemical Composition (wt. pct)

SiC _w Nominal	SiC _w vol. %	Cu	Mg	Li	Fe	Mn
5	6.4	1.04	0.69	0.88	0.040	0.011
10	11.1	0.89	0.71	0.82	0.038	0.014

Results and Discussion

Typical aging hardening results are shown in Figures 1 through 3. The aging response at all temperatures was sluggish, that is little or no change in hardness was initially observed, an increase in hardness only being observed at longer aging times. Moreover the aging time required for this increase in hardness decreased with increasing aging temperature, SiC_w content having little effect on the time required for observation of the initial hardness increase. The hardness curves also tended to converge at lower aging temperature, i.e., below 175°C , with increasing aging time. For example, during natural aging the hardness of the 5 v/o reinforced composite actually exceeded that of the 10 v/o reinforced composite at aging times beyond 500 hrs, Figure 1. Finally, overaging was observed at longer aging times for aging temperature above 150°C , Figure 3.

The hardness increases observed with increasing aging time were not accompanied by an appreciable change in electrical conductivity. Indeed, the recorded increase in electrical conductivity changes was minimal until maximum hardness was approached.

This response to aging time, temperature and SiC_w content can be understood by reference to the transmission electron microscopy observations. Quenching of the solution treated SiC_w reinforced Al-Cu-Mg-Li composites resulted in a heavily dislocated substructure, in agreement with previous investigators[4]. In addition selected area diffraction, Figure 4, suggested that the solution treated condition contained a uniform, fine dispersion of δ' . Subsequent aging initially resulted in δ' coarsening, δ precipitation on dislocations, grain boundaries and whisker-matrix interfaces, also being observed, Figure 5. No evidence for either GPB zones or S' precipitation was observed under these aging conditions, that is prior to any change in

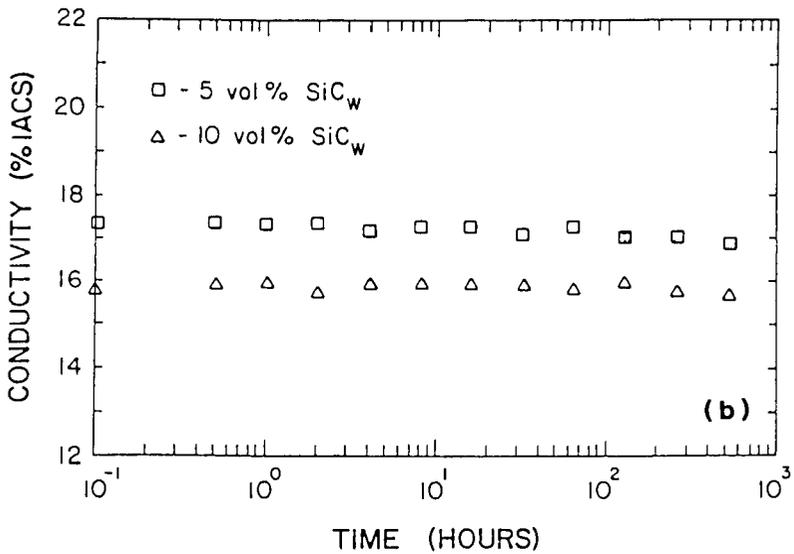
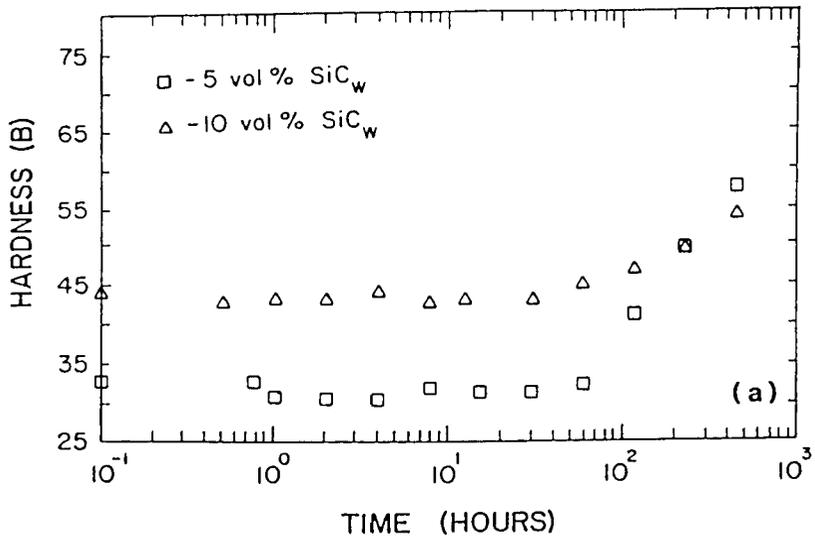


Figure 1: Hardness(a) and electrical conductivity(b) response during natural aging of Al-Cu-Mg-Li reinforced with 5 and 10 v/o SiC_w .

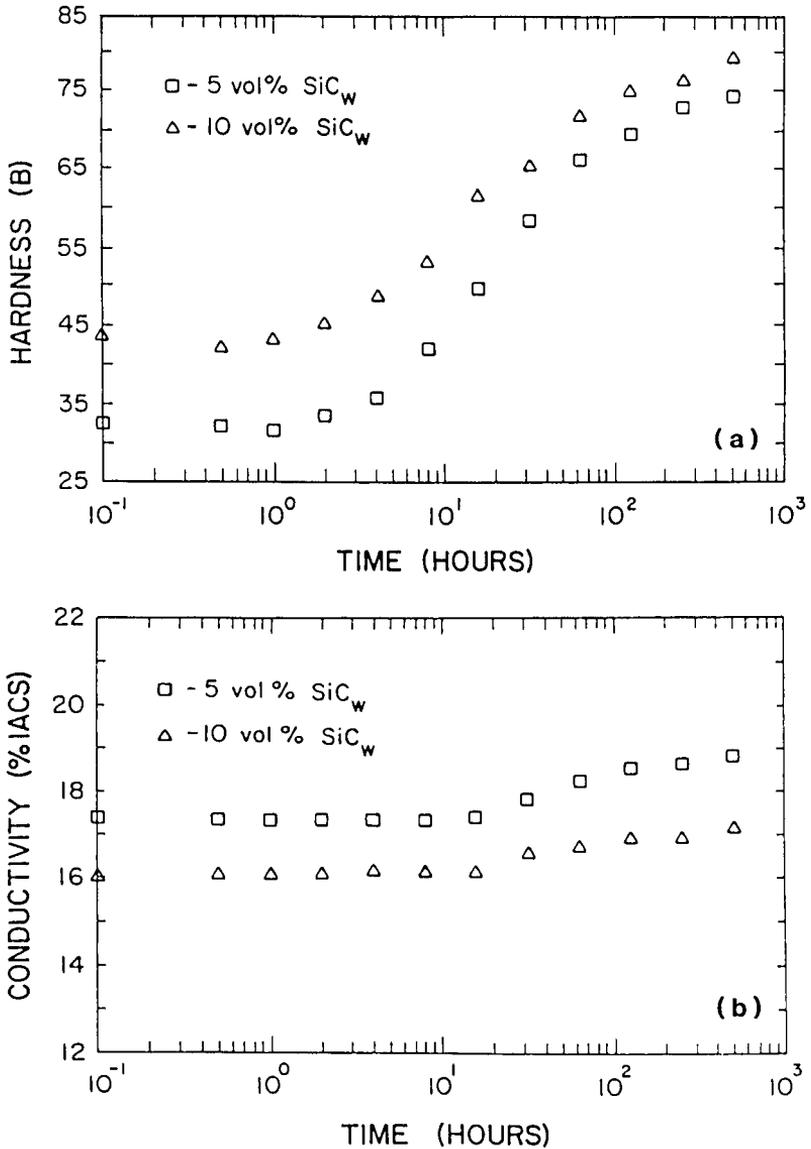


Figure 2: Hardness(a) and electrical conductivity(b) response during aging of Al-Cu-Mg-Li reinforced with 5 and 10 v/o SiC_w at 100°C.

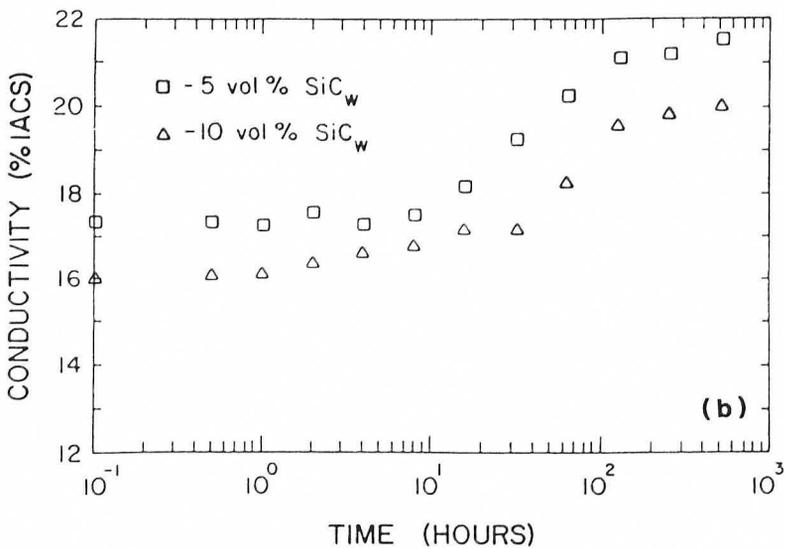
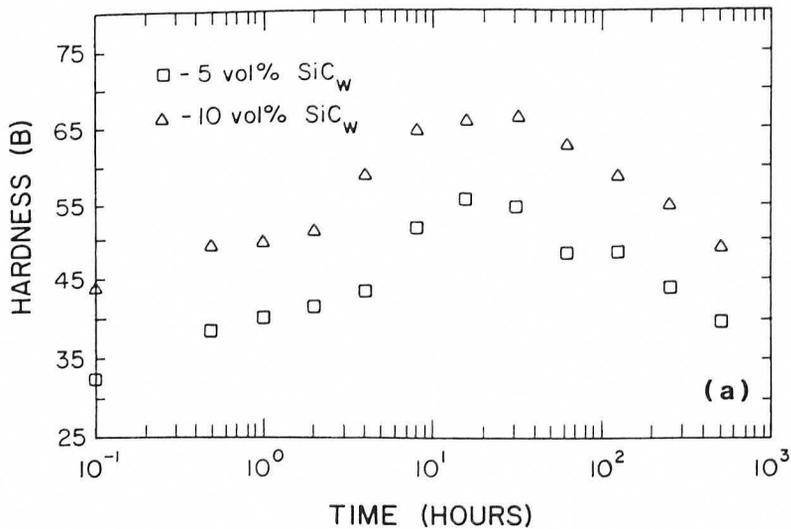


Figure 3: Hardness(a) and electrical conductivity(b) response during aging of Al-Cu-Mg-Li reinforced with 5 and 10 v/o SiC_w at 200°C.

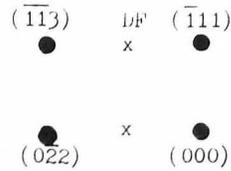
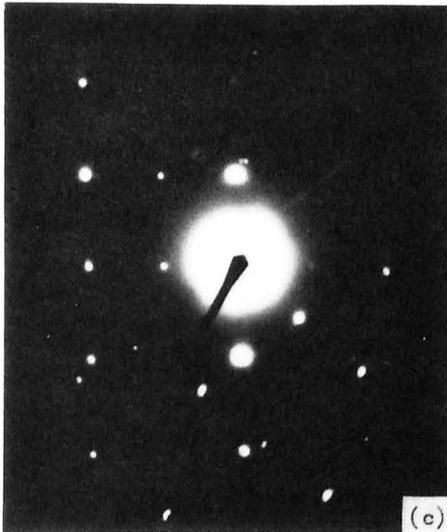
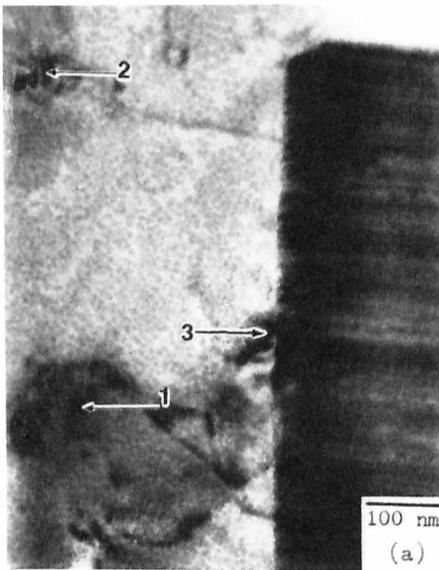
electrical conductivity. Longer aging times, where conductivity increases were observed, did however result in precipitation of S' , Figure 6; again no evidence for GPB formation being observed.

These results suggest that the primary precipitation phenomena controlling the age hardening response of the SiC_w reinforced Al-Cu-Mg-Li alloys examined in this investigation was associated with coarsening of δ' , and to a lesser extent heterogeneous precipitation of δ and S' . This suggestion is supported by the electrical conductivity measurements, no increase in conductivity being observed during initial age hardening. SiC_w content should have little effect on δ' coarsening, indeed the time observed for the initial hardness increase was not a function of SiC_w content. However SiC_w content may influence the volume fraction of δ' formed on quenching from the solution treatment temperature; this effect being manifested by the ability of SiC_w interfaces to serve as effective sinks for vacancies[5]. Increasing SiC_w content is then expected to decrease the vacancy supersaturation maintained during quenching, thereby decreasing the volume fraction of δ' present in the solution treated state. Subsequent aging of a higher volume percent SiC_w reinforced Al-Cu-Mg-Li composite would therefore be expected to result in a lesser hardness increase, this response being observed during natural aging.

Finally, in analogy to other studies of SiC_w reinforced Al-Cu-Mg composites[6], SiC_w reinforcement should decrease the volume fraction of GPB zone formation, again no evidence for their occurrence was observed in these composites. However SiC_w should enhance the δ and S' precipitation and growth, since the formation of both are aided by the dislocation network formed during quenching of SiC_w reinforced composites.



Figure 4: Selected area diffraction pattern illustrating presence of δ' in solution treated Al-Cu-Mg-Li reinforced with 5 v/o SiC_w .



(d)

Figure 5: Transmission electron micrograph of Al-Cu-Mg-Li reinforced with 5 v/o SiC_w aged for 16 hrs at 100°C illustrating fine uniformly dispersed δ' and δ precipitation, the latter on dislocations(1), grain boundaries(2) and SiC_w-matrix interfaces(3). (a) Bright field, (b) dark field, (c) SAD pattern, ZA = (211) and (d) SAD schematic.



Figure 6: Transmission electron micrograph illustrating S' precipitation in Al-Cu-Mg-Li reinforced with 5 aged for 512 hrs at 100°C.

Acknowledgements

The authors wish to acknowledge the financial support of the Lockheed Aeronautical Corporation.

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