Design and Development of an Experimental Wrought Aluminum Alloy for Use at Elevated Temperatures

I. J. POLMEAR and M. J. COUPER

A new wrought aluminum alloy has been designed having high room temperature strength (*e.g.*, 0.2 pct P.S. 520 MPa) combined with improved creep resistance at temperatures in the range 150° to 220 °C. The alloy is Al-6.3 pct Cu-0.5 pct Mg-0.5 pct Ag-0.5 pct Mn-0.2 pct Zr and it is hardened by a new precipitate which forms on the {111} planes and appears to be highly stable at elevated temperatures. Details are given of the principles underlying the development of the alloy and of the preliminary assessment that has been made of mechanical properties. The alloy, or compositions close to it, also has potential for welded applications.

I. INTRODUCTION

AGED aluminum alloys have the highest specific strengths of all commercially available engineering alloys at temperatures up to 100 °C. Between 100 and 200 °C, however, mechanical properties decline rapidly due mainly to coarsening of the fine precipitates on which the alloys depend for their strength.

One approach to improve the elevated temperature performance of aluminum alloys has been the use of rapid solidification technology to produce powders or foils containing high supersaturations of elements such as iron or chromium that diffuse slowly in solid aluminum. In this regard, several experimental materials are now available which have promising creep properties up to 350 °C. Clearly, techniques of this kind will be needed if aluminum alloys are required to operate in such temperature regimes.

Alloys produced by rapid solidification processing tend to be relatively costly and more difficult to prepare than wrought materials fabricated from ingots. Accordingly, an attempt has been made to design a new wrought alloy that might combine the high strength at ambient temperatures with good creep performance in the range 150 to 200 °C. This paper outlines the scientific basis for this design and presents results of a preliminary evaluation of the microstructure and properties of the alloy in the form of extruded bar.

II. EXISTING HIGH-STRENGTH ALUMINUM ALLOYS

The 7000 series of age hardenable alloys that are based on the Al-Zn-Mg-Cu system develop the highest room temperature tensile properties of any aluminum alloys produced from conventionally cast ingots. Values of 0.2 pct proof stress and tensile strength normally exceed 500 MPa and 550 MPa if the alloys are aged to the T6 condition. However, the strength of these alloys declines rapidly if they are exposed to temperatures exceeding 100 to 120 °C (Figure 1).[1] Alloys of the 2000 series such as 2014 (Al-4.3 pct Cu-0.5 pct Mg-0.8 pct Si-0.6 pct Mn)* and 2024 (Al-4.3 pct Cu-1.5 pct Mg-0.6 pct Mn) perform better above these temperatures but are not normally used for elevated temperature applications.

One wrought aluminum alloy commonly favored for elevated temperature applications is 2618 (Al-2.4 pct Cu-1.5 pct Mg-1 pct Fe-1 pct Ni-0.2 pct Si) and it has particular application in forgings. Typical room temperature tensile properties of 2618 in the T6 condition are 0.2 pct proof stress: 380 MPa, tensile strength: 440 MPa. The mechanical properties of 2618 become superior to the 7000 alloys above 120 °C (Figure 1). For temperatures exceeding approximately 220 °C, however, another 2000 series alloy 2219 (Al-6.3 pct Cu-0.3 pct Mn-0.18 pct Zr-0.1 pct V) displays greater mechanical strength although its room temperature properties (0.2 pct proof stress: 290 MPa, tensile strength: 415 MPa for T6 condition) are significantly lower than 2618. 2219 has the advantage of being weldable.

Alloy 2618 is based on the Al-Cu-Mg system and it derives its mechanical properties from a combination of precipitation and dispersion strengthening. The main precipitates are GP(Cu, Mg) zones (also known as GPB zones)[2] which form rapidly when the alloy is aged at room temperature, and almost immediately at temperatures up to at least 200 °C,[3] together with the semi-coherent phase S' which is

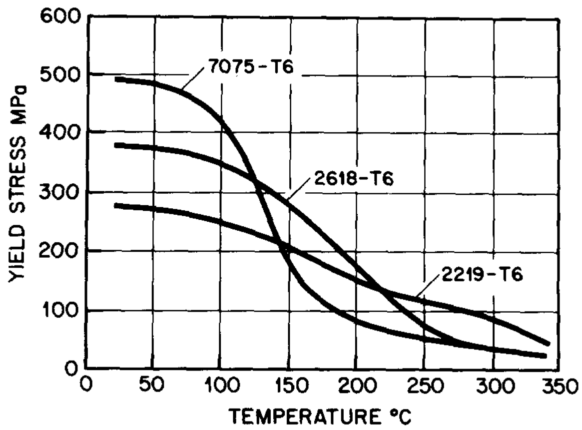

Fig. 1 — Values of 0.2 pct proof stress of aluminum alloys after exposure for 1000 h at temperatures between 0 and 350 °C.

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* Unless stated otherwise, alloy compositions and additions are expressed in percentages by weight.

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Manuscript submitted January 20, 1987.

orthorhombic. The S' phase forms on the {210} matrix planes and it may be dispersed uniformly as rods parallel to the ⟨100⟩ directions of the matrix, or as corrugated laths or sheets that nucleate on dislocation lines. The presence of stable particles of the phase Al_9FeNi serves to control grain size.[4]

2219 is essentially an Al-Cu alloy hardened on artificial aging at 190 °C by the partly coherent phase θ' which is tetragonal and forms on the {100} matrix planes.[5] Some dispersion strengthening will occur through the formation of compounds containing manganese and zirconium that control grain structure whereas the role of vanadium seems uncertain. The fact that 2219 displays superior mechanical properties to 2618 at the upper end of the elevated temperature range implies that the precipitate θ' is more stable than S', and mechanical property data reported by Chellman[6] support this conclusion.

III. TRACE ELEMENT EFFECTS OF SILVER IN 2000 SERIES ALLOYS

It has been shown that small amounts (~0.1 at. pct) of silver may modify precipitation processes in aluminum alloys that contain magnesium.[3,7] Thus, although silver has no significant effect in artificially aged Al-Cu alloys,[3] it may completely change the precipitation processes that normally occur in aged Al-Cu-Mg alloys.[3,7–15]

For most compositions in the Al-Cu-Mg system that are aged in the temperature range of 100° to 250 °C, silver both increases the component of hardening derived from the formation of GP (Cu, Mg) zones and stimulates precipitation of a finely dispersed, cubic T phase ($Al_6(Cu, Ag)Mg_4$) rather than S'.[8,9,10] In this regard, the T phase appears to form despite the fact that the alloys lie in the $\alpha + S$ region of the ternary phase diagram at the aging temperature (Figure 2). The silver-containing alloys show an enhanced response to age hardening with a consequent improvement in both room and elevated temperature tensile properties. However, the T phase was found to be metastable, and prolonged aging (*e.g.*, 100 days at 200 °C) leads to its resolution and replacement by S' (or S) so that the properties revert to those of the silver-free alloys.[9]

For ternary alloys with high Cu:Mg ratios such that they lie in the $\alpha + \theta + S$ and possibly $\alpha + \theta$ regions of the Al-Cu-Mg phase diagram, aging normally leads to precipitation of phases that form in both the Al-Cu and Al-Cu-Mg systems.[9] However, if silver is added to these alloys, there is still a marked increase in age hardening[9] and it has been shown that a new phase becomes the dominant precipitate on aging above 100 °C.[11–18] This phase, which has been designated Ω, forms on the {111} matrix planes as a uniform dispersion of large, but very thin hexagonal-shaped plates and its appearance is shown in Figure 3. There is strong evidence that Ω is coherent with the matrix at the ends of the plates (Figure 4). There is debate as to whether Ω is a monoclinic,[16] hexagonal,[17] or orthorhombic[18] form of the phase $CuAl_2$, which is itself body-centered tetragonal. Moreover, no explanation has so far been proposed to account for the way that small amounts of magnesium and silver can be incorporated into the structure.

The characteristics of the Ω phase have been studied in detail for the alloy Al-4 pct Cu-0.3 pct Mg-0.4 pct Ag.[13,14,15] Nucleation is an energetically easy process and the mechanism appears to be fundamentally different from that involved with the phases θ', θ, and S. Ω may be an equilibrium precipitate and there are several observations which suggest that it is inherently more stable and resistant to coarsening than θ'. For example, comparisons of the creep performance of sheet specimens of the above alloy with other Al-Cu-Mg alloys have supported this conclusion.[18] It has also been found that Ω may coexist with other precipitates such as θ'.[15]

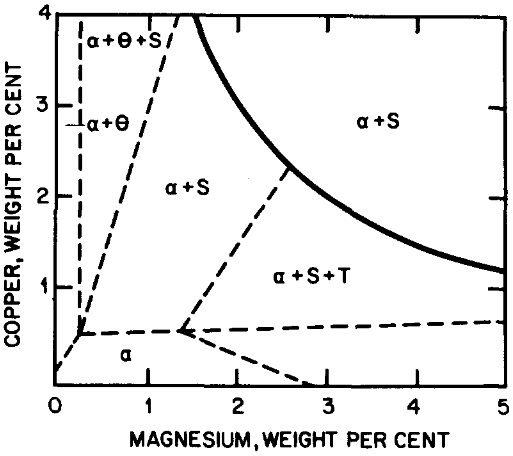
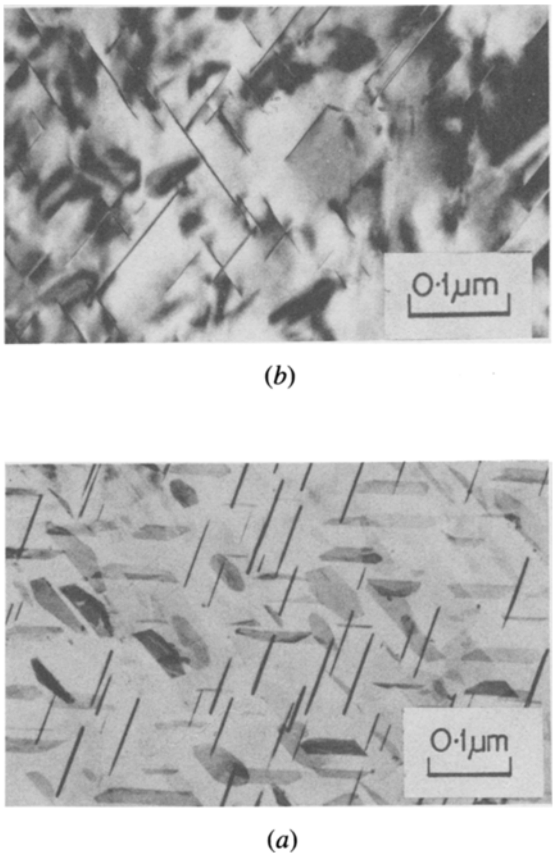


Fig. 2—Phases present in the Al-Cu-Mg system at 500 °C (solid line) and 190 °C (dotted line).



(*b*)

(*a*)

Fig. 3—Thin, hexagonal shaped plates of the Ω phase precipitated on the {111} planes in the alloy Al-4 pct Cu-0.3 pct Mg-0.4 pct Ag aged to peak hardness at 200 °C. (Courtesy (*a*) R. Chester and (*b*) B. Muddle)

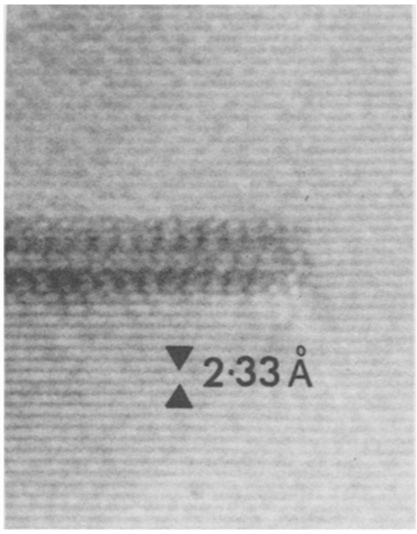


Fig. 4—Lattice image of end of an Ω plate showing apparent coherency with matrix planes. (Courtesy K. Knowles)

It should be noted that there is a commercial Al-Cu-Mg-Ag casting alloy containing a somewhat larger amount of silver. This has been designated A201 and it has the nominal composition Al-4.3 pct Cu-0.35 pct Mg-0.7 pct Ag-0.45 pct Mn. It was developed in the United States[20] and there is a similar alloy available in France which is known as Avior and has, in addition, 1.3 pct zinc.[21] As premium quality castings, these alloys are quoted as giving guaranteed minimum properties of 0.2 pct proof stress: 350 MPa, tensile strength: 415 MPa in the T6 temper (commonly 16 hours at 155 °C), although significantly higher values have been recorded. Misra and Oswalt[22] have attributed the high properties of A201 to the effect of silver in stimulating the aging process that normally occurs in Al-Cu alloys (*i.e.*, GP (Cu) zones $\rightarrow \theta'' \rightarrow \theta' \rightarrow \theta$). However, it is clear that this is not so as Ω is the dominant precipitate in this[23] and similar compositions.[11-18]

IV. DESIGN OF A NEW CREEP-RESISTANT WROUGHT ALUMINUM ALLOY

The foregoing discussion, together with certain practical requirements, suggests that the following criteria are relevant in attempting to predict a new alloy which may combine high strength at room temperature with improved creep resistance at temperatures in the range 150 °C to 200 °C:

1. The alloy should be based on the Al-Cu system. 2219 is one such alloy but its tensile properties at room temperature are far too low.
2. The alloy should contain small additions of silver and magnesium so that artificial aging promotes formation of the relatively stable phase Ω. This phase may coexist with θ' which also has good elevated temperature stability. The silver content should be as low as possible because of the cost of this addition.
3. It will be necessary to add elements that promote the formation of fine, stable intermetallic compounds that control grain structure.
4. In common with other wrought alloys, the new alloy must show adequate ductility at room and elevated temperatures (*e.g.*, >5 pct in a standard tensile test).
5. It would be desirable if the alloy could be capable of being quenched into boiling water after solution treatment so as to minimize residual stresses.

Selection of Alloy Composition

1. The alloy Al-4 pct Cu-0.3 pct Mg-0.4 pct Ag, which has already been studied, meets the first two criteria although it has inadequate mechanical strength in its present composition. It was therefore decided to raise the copper content to the level present in alloy 2219 (6.3 pct) and that proportionate increases be made in the levels of magnesium and silver. Thus it was proposed that the age hardening component of the new alloy be Al-6.3 pct Cu-0.5 pct Mg-0.5 pct Ag.

2. Several transition metals combine with aluminum to give stable dispersoids which cause both dispersion strengthening and grain refinement. By analogy with 2618, it could be suggested that additions of 1 pct iron and 1 pct nickel be made to the experimental alloy since balanced amounts of these two elements can lead to formation of the compound Al_9FeNi. However, in view of the high copper content, iron may tend to interact with at least some of this solute forming the compound Al_7Cu_2Fe, thereby reducing the amount of copper available to contribute to precipitation hardening. Instead, it was proposed to limit the iron content to a maximum of 0.15 pct and to omit nickel. Manganese in the range of 0.4 to 0.8 pct is another transition metal commonly added to aluminum alloys and it has the additional advantage of combining with iron to form the compound $Al_6(Mn, Fe)$. Zirconium (0.10 to 0.25 pct) is also now added to many high-strength aluminum alloys because it may be precipitated as finely dispersed, stable particles of Al_3Zr. It was proposed, therefore, that the alloy should contain 0.5 pct manganese and 0.2 pct zirconium.

3. Silicon may combine preferentially with magnesium to form Mg_2Si, thereby reducing the interaction of this element with silver. Moreover, silicon combines with other elements to form eutectics which lower the temperature at which liquation occurs during solution treatment. Accordingly, it was proposed to limit silicon to a maximum of 0.10 pct.

4. Titanium (0.05 pct) was added to promote grain refinement. Thus the complete composition proposed for the experimental alloy was as follows: Al-6.3 pct Cu-0.5 pct Mg-0.5 pct Ag-0.5 pct Mn-0.2 pct Zr-0.05 pct Ti with iron and silicon a maximum of 0.15 pct and 0.10 pct, respectively.* It may be noted that, as with 2219, it was expected that the experimental alloy would be weldable.

V. PRELIMINARY EVALUATION

A small ingot of the proposed alloy was chill cast and homogenized for 24 hours at 480 °C (designated Alloy A). This had the composition Al-6.7 pct Cu-0.46 pct Mg-0.50 pct Ag-0.48 pct Mn-0.18 pct Zr-0.07 pct Si-0.12 pct Fe-

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*Alloys based on this composition have been the subject of patent applications.

0.06 pct Ti. Although the copper content was higher than planned, cylinders 36 mm diameter were machined from the ingot and successfully extruded at 420 °C to produce 9 mm diameter bar (extrusion ratio 13:1). Part of the bar was used to establish the optimum heat treatment cycle and for studying the precipitation process by electron microscopy and differential thermal analysis. The remainder provided specimens for measuring tensile properties at ambient and elevated temperatures and for determining stress-rupture data from creep tests conducted at 180 °C and 220 °C.

A. *Heat Treatment Cycle*

Solution Treatment. Because of the high content of alloying elements, especially copper, it seemed necessary that the solution treatment temperature be as high as possible. Sections of the extruded bar were exposed for 6 hours at temperatures in the range 520 to 540 °C and microscopic examination revealed that the first sign of liquation (overheating) occurred at 530 °C. Consequently, a temperature of 525 °C was selected for solution treatment.

Quenching. Results were obtained for specimens that were quenched in either cold water or boiling water prior to aging. In the latter case, the specimens were immersed for a standard time of 60 seconds.

Aging. The optimum aging conditions were determined by preparing a series of hardness/time curves for specimens aged at 175 °C, 185 °C, 195 °C, and 205 °C (Figure 5). The peak hardness was similar in each case (165 to 167 H_B), which is not commonly observed for such a wide range of temperatures. The curves were characteristically rather flat in this region. Aging times to achieve maximum hardening ranged from 20 to 50 hours at 175 °C to 1.5–2.5 hours at 205 °C.

For the present work, a standard aging time of 7 hours at 195 °C was selected and complete hardness/time curves for cold water and boiling water quenched specimens aged at 195° are shown in Figure 6. Curves determined in an earlier study of the alloy 2618[9] have been included for comparison. It will be noted that the alloy A aged at a much faster rate than 2618 and hardening was also significantly greater.

Experiments also indicated that, unlike some other aluminum alloys, the response of alloy A to artificial aging was not affected by delays at room temperature after quenching, or by the rate of heating to the aging temperature.

B. *Microstructure*

After solution treatment, the extruded bar contained relatively coarse, elongated grains which themselves comprised well-defined, equiaxed subgrains with an average size of 5 μm. Nonuniform dispersions of undissolved intermetallic compounds were present which are attributed mainly to the fact that the copper content was considerably in excess of the solid solubility of this element. Transmission electron microscopy revealed that precipitation hardening was due to the presence of a fine and uniform dispersion of the plates of the Ω phase throughout the grains (Figure 7).

C. *Mechanical Properties*

Tensile Properties. The tensile properties for alloy A at 20° and 200° are shown in Table I where they may be compared with typical values for several commercial alloys.[1] Relationships of 0.2 pct proof stress with temperature for alloy A and the alloys 7091 and 2618 are also shown in Figure 8. It will be noted that the room temperature tensile properties of alloy A were greater than those of the other wrought alloys including 7075-T6 and that this superiority was maintained in tests carried out at 200 °C. It was also found that, above 125 °C, alloy A had a higher elevated

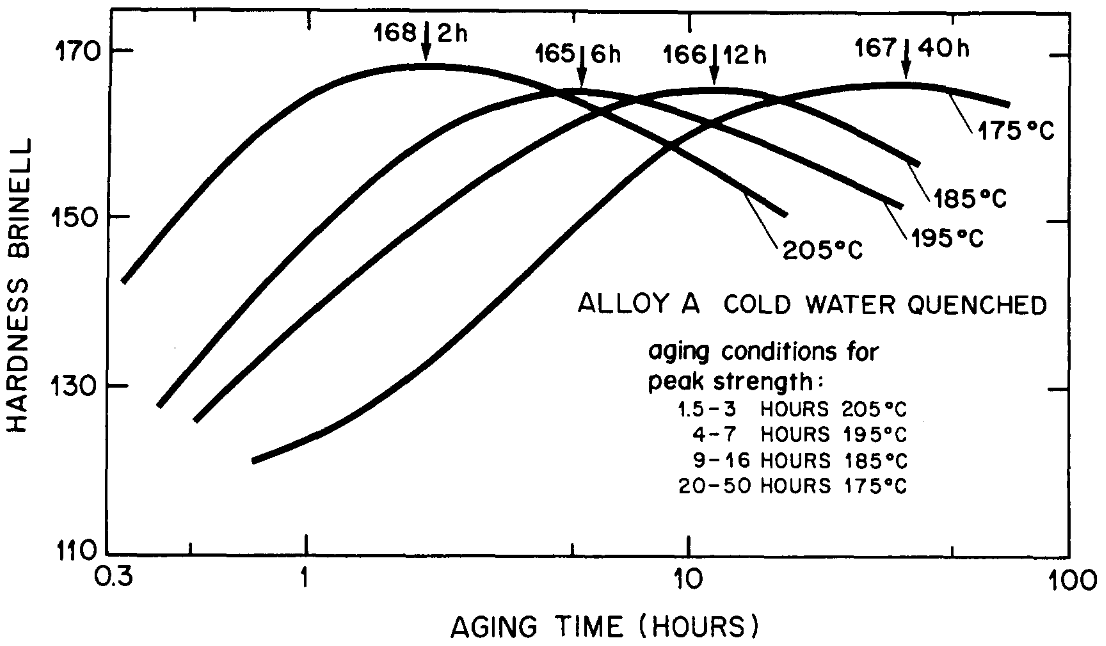

Fig. 5—Hardness-time curves for alloy A aged at 175, 185, 195, and 205 °C.

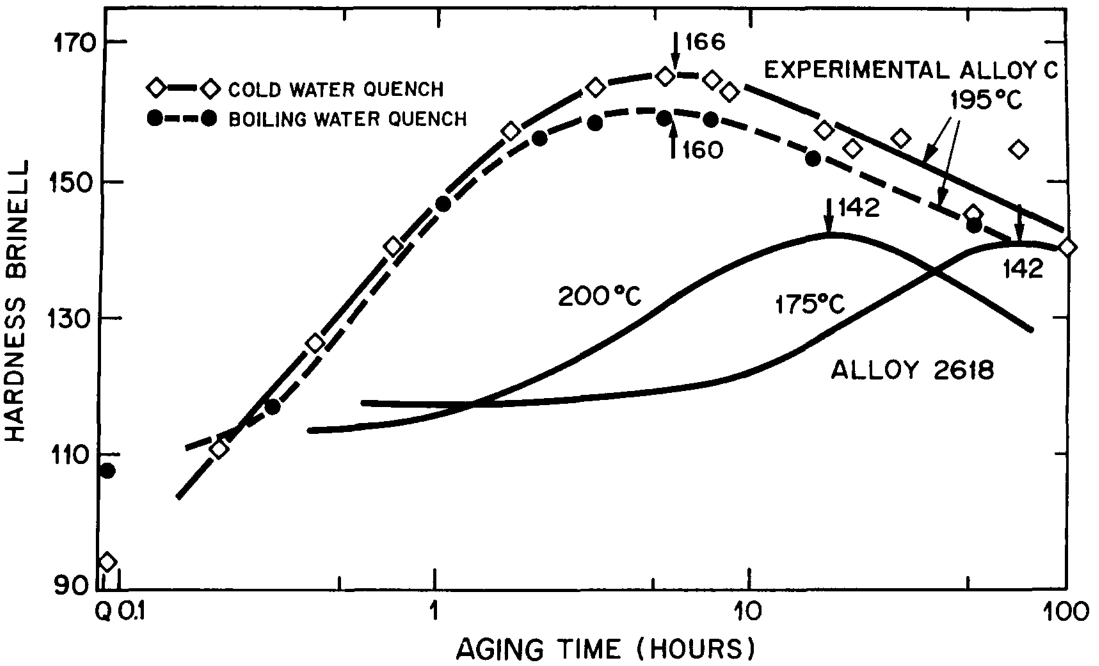

Fig. 6—Hardness-time curves for alloy A aged at 195 °C after (*a*) cold water quenching (T6) and (*b*) boiling water quenching (T64). Curves for cold water quenched specimens of alloy 2618 are also included.

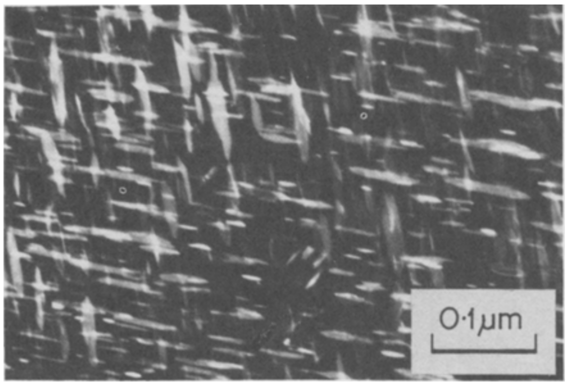

Fig. 7—Dark-field transmission electron micrograph showing plates of Ω phase on {111} planes of alloy A aged 7 h at 195 °C.

temperature strength than 7091 which is produced by powder metallurgy techniques.

Boiling water quenching alloy A prior to aging at 195 °C caused reduction of approximately 5 pct in values at 0.2 pct proof stress as compared with quenching into cold water (Table I). As mentioned above, a delay at room temperature prior to aging at 195 °C had no significant effect on tensile properties. For example, the 0.2 pct proof stress of the alloy after a delay of 3 days was 519 MPa which compares with the average value of 517 MPa that was recorded from specimens aged immediately after quenching.

Stress-Rupture Properties. An indication of the likely creep behavior of alloy A was obtained by determining stress-rupture values at 180 °C and comparing them with those reported for the 2000 series alloys. As shown in Figure 9, alloy A performed significantly better, particu-

Table I. Tensile Properties of Commercial Aluminum Alloys and the Experimental Alloy A at 20 and 200 °C

	Test Condition									
	20 °C			0.5 Hour, 200 °C (1)			100 Hours, 200 °C (1)			
Alloy	0.2 Pct P. S. MPa	T. S. MPa	Pct Elong.	0.2 Pct P. S. MPa	T. S. MPa	Pct Elong.	0.2 Pct P. S. MPa	T. S. MPa	Pct Elong.	Remarks
7075-T6	495	565	11	235	285	12	125	140	16	extrus.
2618-T6	385	445	7.5	310	350	9	270	290	12	extrus.
2219-T6	295	420	10	235	295	16	225	275	20	extrus.
2021-T8	435	500	10	275	320	16	240	305	20	plate
Expt. Alloy A										
(2) T6	517	581	7.5	393	409	11	305	325	18	extrus.
(3) T64	490	547	7.5	376	381	15.5	—	—	—	extrus.

(1) Tests carried out at 200 °C.
(2) The aging treatment was 7 hours at 195 °C.
(3) T64 indicates boiling water quenching.

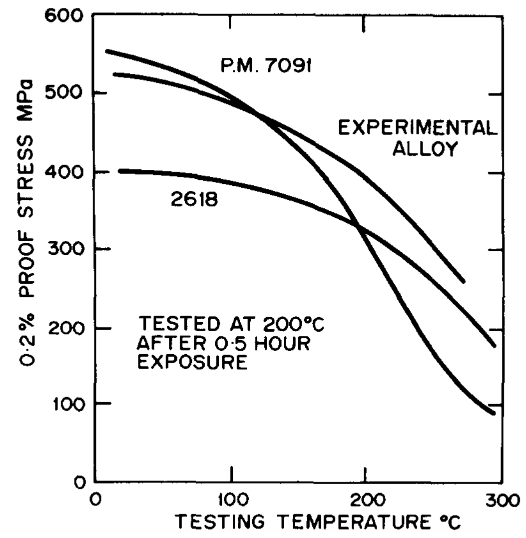

Fig. 8 — Values of 0.2 pct proof stress in the range 20 to 300 °C for alloy A, 2618, and the powder metallurgy alloy 7091.

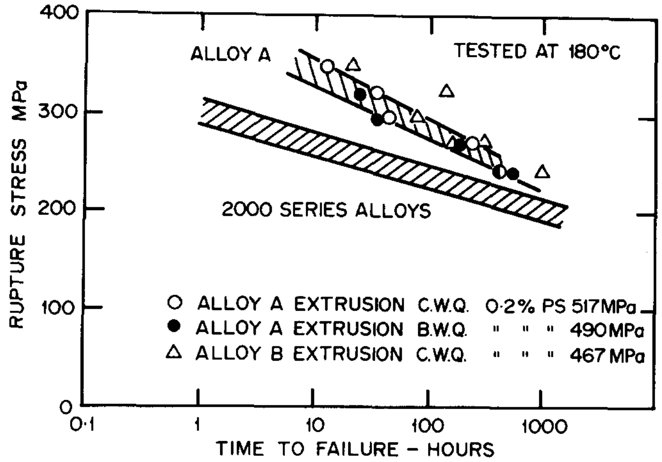

Fig. 9 — Stress-rupture results for creep tests at 180 °C on alloys A and B compared with those for 2000 series alloys.

larly at high stress levels, and average lives were more than an order of magnitude longer over most of the stress range that was tested. Tests at 220 °C also showed that alloy A was superior to results reported for 2000 series alloys.

VI. SECOND STAGE EVALUATION

A. *Optimization of Alloy Composition*

Age Hardening Response. It was necessary to determine whether the predicted alloy composition was in fact the one which would give the greatest response to age hardening. Moreover, since silver is a costly addition, there was a need to ascertain the minimum amount required to cause the desired changes in microstructure and mechanical properties.

These requirements were satisfied by preparing small quantities of 14 alloys containing two levels of copper (5.3 and 6.3 pct) with magnesium and silver levels each in the range 0 to 0.5 pct. In each case, the response of the alloy to age hardening at 195 °C was determined and sections of some of the hardness/time curves are shown in Figures 10 and 11.

Figure 10 includes the aging curve for the binary alloy Al-5.3 pct Cu and it will be seen that the addition of 0.4 pct silver alone actually caused a slight decrease in the response to hardening. Progressive increases in hardening occurred on adding 0.4 pct magnesium and 0.4 pct each of this element and silver. Raising the level of magnesium to 0.5 pct in the presence of 0.4 pct silver caused an additional significant increase in the response to aging and the peak hardness value at 195 °C changed from 137 to 147. Raising the level of copper to 6.3 pct increased the response of comparable alloys to age hardening (Figure 11). If a peak hardness of 150 is proposed as a target, then alloys containing 6.3 pct Cu-0.45 and 0.5 pct Mg-0.4 pct Ag and 6.3 pct Cu-0.5 pct Mg-0.3 to 0.5 pct Ag meet this value. The alloy Al-5.3 pct Cu-0.5 pct Mg-0.4 pct Ag (peak hardness 147) is close to it.

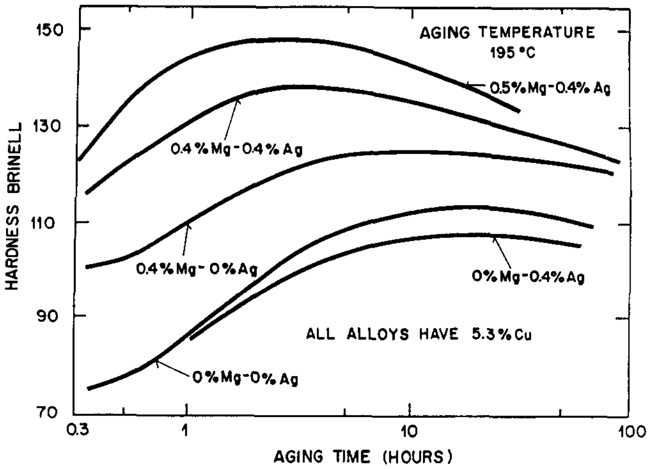

Fig. 10 — Hardness-time curves at 195 °C for alloys based on Al-5.3 pct Cu.

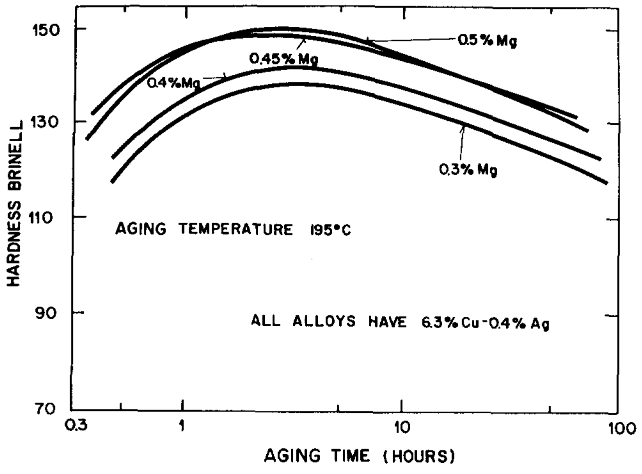

Fig. 11 — Hardness-time curves at 195 °C for alloys based on Al-6.3 pct Cu-0.4 pct Ag.

So far as the minimum silver content is concerned, it seems possible to reduce this to at least 0.3 pct if a peak hardness of 150 or more is required (Table II).

From all these results it is proposed that the composition of the alloy with respect to those elements which contribute to age hardening should lie within the range Al-6.0 pct to 6.5 pct Cu-0.45 to 0.5 pct magnesium and 0.3 to 0.4 pct silver which is, in fact, close to that predicted.

Grain Size Control. Additions of 0.5 pct manganese and 0.2 pct zirconium seemed to provide fine grain size and caused little removal of copper into intermetallic compounds in alloy A so it was decided to prepare a second alloy (B) with these elements present. As mentioned earlier, the possibility of adding 1 pct iron and 1 pct nickel was considered and it was decided to prepare a second alloy containing these elements (Alloy C).

Impurities. It was again decided to limit the silicon content to a maximum of 0.10 pct. In alloy B a limit of 0.15 pct was also set on the iron content.

B. *Preparation of Alloys*

Melts of approximately 15 kg of alloys B and C were prepared and chill cast as cylinders 60 mm in diameter. The castings were then given a multi-stage homogenization treatment comprising heating to 450 °C, heating at 20 °C/hr to 475 °C, holding for 6 hours, heating 10 °C/h to 500 °C, holding for 18 hours, followed by a slow cool to room temperature.

Compositions were as follows:

Alloy B: Al-6.0 pct Cu-0.45 pct Mg-0.5 pct Ag-0.5 pct Mn-0.14 pct Zr
(Si0.07 pct, Fe0.12 pct)
Alloy C: Al-6.0 pct Cu-0.45 pct Mg-0.5 pct Ag-1 pct Fe-1 pct Ni
(Si0.07 pct)

Alloy B is similar to alloy A but with a lower copper content while alloy C incorporates iron and nickel rather than manganese and zirconium for dispersion strengthening and grain size control.

Table II. Peak Hardness Values for High-Purity Alloys Aged at 195 °C

5.3 Pct Cu

Ag \ Mg	0	0.4 Pct	0.5 Pct
0	115	124	
0.4 pct	110	137	147

6.3 Pct Cu

Ag \ Mg	0.3 Pct	0.4 Pct	0.45 Pct	0.5 Pct
0 pct		130		
0.2 pct				146
0.3 pct				153
0.4 pct	138	142	150	152
0.5 pct				153

Hardness in Brinell units.

The cylinders were extruded at 440 °C to produce 11 mm diameter bar and 7 × 22 mm flat sections. Heat treatment was the same as described previously.

C. *Mechanical Properties*

Tensile Properties. Values determined for alloys B and C aged 7 hours at 195 °C are shown in Table III.

The tensile properties of both alloys fell below those recorded for the alloy A produced in the first stage of the evaluation (Table I). This is attributed both to the lower levels of copper in alloys B and C, and to the unexpected loss from the matrix of copper which X-ray energy dispersive microanalysis showed to be combined with manganese, iron, and nickel in intermetallic compounds. This latter effect was particularly apparent in alloy C, indicating that large additions of iron and nickel are undesirable and this alloy was discarded. The greater loss of copper to intermetallic compounds in alloy B than occurred with alloy A seems likely to be due to the more complex homogenization treatment used for alloy B.

D. *Stress-Rupture Properties*

Results for alloy B have also been included in Figure 9. It will be seen that, despite the lower levels of tensile properties, this alloy generally performed rather better than alloy A. For example, one specimen failed after 1016 hours at a stress of 240 MPa which compares with averages of 450 hours for alloy A and 70 hours for the 2000 series alloys. This is attributed to the improved microstructure in alloy B, notably the finer dispersion of intermetallic compounds as compared with alloy A.

E. *Stress-Corrosion Resistance*

A semi-quantitative indication of the susceptibility of the alloy to stress-corrosion cracking was obtained using the "cut-edge" method described by Gruhl and Schippers.[24] This technique involves shearing off pieces from flat sections of alloys which produces internal stresses in the direction normal to the sample surface, and exposing the pieces to corrosive solutions. For the present work, extruded flat sections of alloy B, 7 mm thick, were used and the test solutions were:

1. 2 pct NaCl + 0.5 pct Na$_2$CrO$_4$
   (pH3), room temperature
2. 1 pct NaCl + 2 pct K$_2$Cr$_2$O$_7$
   (pH4), 45 °C

Results are expressed in times taken for the appearance of the first crack in the "cut-edge" surface. These are shown in Table IV and some comparisons are made with results provided for some commercial alloys.[25]

Table III. Tensile Properties of Alloys B and C

Alloy	0.2 Pct P. S. MPa	T. S. MPa	Pct Elong.
B-T6	467	507	10.5
-T64	445	496	10.3
C-T6	333	386	13.0

Table IV. Results of "Cut-Edge" Stress-Corrosion Tests

Alloy	Condition	0.2 Pct P. S. MPa	Time to Failure, Days
Solution 1			
Alloy B	T6	467	60 (1 spec.)
			>60 (8 spec.)
	T64	445	>60 (8 spec.)
7075	T6	523	0.5 to 3
	T73	462	>50
2014	T6	348	6 to 20
2024	T6	367	0.5 to 1
Note: Gruhl and Schippers[23] consider alloys which do not develop cracks in 50 days as being relatively insensitive to stress-corrosion cracking.			
Solution 2			
Alloy B	T6	467	10 (1 spec.)
			25 (4 spec.)
	T64	445	15 (1 spec.)
			>15 (7 spec.)
Note: Failure times exceeding 15 days are considered to indicate a high resistance to stress-corrosion cracking in Solution 2.[23]			

These results, although preliminary in nature, show that alloy B performs significantly better than several commercial alloys and seems likely to show little sensitivity to stress-corrosion cracking.

F. *Weldability*

As mentioned earlier, the commercial alloy 2219, which is basically Al-6.3 pct Cu, is weldable and preliminary trials with alloy B have shown that it, too, can be welded. More work is needed to assess the mechanical properties of selected welded joints, but the microstructure suggests the welds are of good quality (*e.g.*, hot cracking was not observed).

VII. FURTHER WORK

The results so far obtained on alloys A and B seem sufficiently encouraging to justify the production of a larger ingot so that a more complete assessment of properties can be obtained. For example, creep behavior at temperatures in the range 150° to 220 °C will need to be determined in more detail. Another requirement will be the production of larger wrought sections which will allow the fracture toughness and weldability of the alloy to be assessed.

So far as composition is concerned, it would be desirable to produce and test alloys with lower silver contents (*e.g.*, 0.25 and 0.3 pct), as it seems likely that the cost penalty associated with this addition can be reduced.

VIII. CONCLUSIONS

1. An experimental wrought alloy based on the predicted composition Al-6.3 pct Cu-0.5 pct Mg-0.5 pct Ag-0.5 pct Mn-0.2 pct Zr has exhibited a combination of tensile properties at room temperature and 200 °C that exceed those of existing commercial aluminum alloys.
2. The superior mechanical properties of the new alloy are attributed to the presence of a fine and uniform dispersion of the relatively stable precipitate, designated Ω, that forms over the $\{111\}$ planes during aging.
3. The stress/rupture properties of the experimental alloy are significantly better than those for the 2000 series of alloys at temperatures of 180 °C and 220 °C.
4. The alloy appears to be relatively insensitive to stress-corrosion cracking and is significantly less susceptible than several commercial aluminum alloys.
5. The alloy shows relatively little quench sensitivity and offers considerable flexibility in aging conditions.
6. Room and elevated temperature ductility is adequate.
7. As with 2219, the experimental alloy is capable of being welded.
8. It is probable that cost savings can be achieved by reducing the level of silver to around 0.3 pct.

ACKNOWLEDGMENTS

Most of the work described in the paper was carried out when Professor Polmear was on sabbatical leave from Monash University and attached to the BBC Brown Boveri Research Centre, Baden-Dättwil, Switzerland, from July to November 1985. Grateful acknowledgment is made of useful discussions with Dr. R. Singer and for experimental contributions made by R. Baumann, E. Schönfeld, Frau R. Sebalj, and Frau V. Schmid. Alloys B and C were cast and extruded at Alusuisse Research Laboratories, Neuhausen, Switzerland, through the courtesy of Dr. P. Furrer and Mr. G. Höllrigl. Dr. H. Cordier of the Leichtmetall-Forschungsinstitut, Vereinigte Aluminium-Werke, Bonn, kindly arranged the stress-corrosion tests. Weldability trials were arranged at the Comalco Research Centre, Melbourne, through Dr. T. Sritharan.

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