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This study determined the independent effects of various homogenization cycles and precipitation treatments on the elevated temperature workability and the final ambient temperature mechanical properties of AA2024 aluminum alloy and on the T3 tensile properties of 2014 aluminum alloy as well as T6 tensile properties of 2618 and 2618 (Cu-rich) aluminum alloys. The elevated-temperature tensile and extrusion tests indicate that the workability of AA2024 alloy improves with elevated-temperature precipitation treatment as suggested by earlier investigations. The precipitation treatments do not appear to degrade the ambient-temperature T3 and T8 tensile properties. The time at the precipitation temperature appears to affect the T3 and T8 tensile properties in unextruded ingot, longer times especially providing both relatively high ambient-temperature strength and ductility of AA2024 alloy. The time at the standard homogenization temperature and the heat-up and cool-down rates do not dramatically affect the T3 tensile properties of unextruded ingot of AA2024 and 2014 alloys. However, long soak times at the homogenization temperature

and more rapid cooling rates may improve the properties somewhat of AA2024 alloy and longer heat-up times and rapid cooling rates may slightly improve the properties of 2014 alloy. The higher standard solution temperature appears to increase both strength and ductility of 2014 alloy over lower temperatures. The homogenization temperature affects the T6 tensile properties of 2618 and 2618 (Cu-rich) alloys, a high homogenization temperature (compare to standard homogenization temperature) providing both high strength and ductility. Increased manganese and copper appears to increase the strength, but slightly decreases the ductility. The standard aging temperature and time produce higher strength but lower the ductility than lower temperatures at the same or shorter aging times in 2618 (Cu-rich) alloy.

The Effects of Thermal Processing on the Mechanical Properties of AA2024, 2014 and 2618 Aluminum Alloys

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INTRODUCTION

A unique combination of properties makes aluminum and its alloys the second most widely used material, after iron and its alloys, in the world. 2xxx aluminum alloys, possessing the common good characters such as high strength-weight ratio, good electrical and thermal conductivities, less toxic reaction, high resistance to corrosion and good workability, become one of the most important commercial alloys in aluminum industry. 2xxx aluminum alloys consisting of copper (principal alloy element) and other additional elements require heat-treatment to obtain optimum properties. Fig.1 illustrates the common production process of 2xxx aluminum alloys [1,2,3]. Among them, homogenization and heat treatment are main factors that affect the thermal and mechanical properties of the material [4,5].

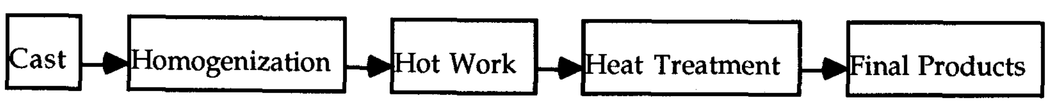


Fig.1 The production process of 2xxx Al alloys

Homogenization. The initial thermal operation applied to ingots prior to hot working is referred to as "ingot preheating" or "homogenization" and has one or more purposes depending on the alloy, product, and fabricating process involved. It consists of three steps: heating up to homogenization temperature, a soak at the homogenization temperature, and cooling down to low temperature (e.g. ambient temperature). Its principal objective for 2xxx aluminum alloys is to improve hot workability by eliminating the initial cast structures. The microstructure of aluminum-copper alloys in the as-cast condition is a cored dendritic structure with solute content increasing progressively from center to edge with an interdendritic distribution of second-phase particles or eutectic. Because of the relatively low ductility of the intergranular and interdendritic networks of these second-phase particles, as-cast structures generally have inferior workability and heterogeneous distribution of alloying elements. Solution of the intermetallic phases rejected interdendritically during solidification by the homogenization operation is only one step toward providing maximum workability. Because most of the solute is in solid solution after heating and soaking, further softening and improvement in workability can be obtained by slow cooling, to reprecipitate and coalesce the solute as fairly large particles. Greatly extended homogenizing periods result in a higher rate of extrusion and in an improved surface appearance of extruded products. Therefore, the soaking temperature and time as well as cooling rate greatly affect the workability of 2xxx aluminum alloys [5]. Although the study of homogenization is very important, the method used to study homogenization of cast structures for improved workability were developed chiefly by empirical methods, correlated with optical metallographic examinations, to determine the time and temperature required to minimize coring and dissolve the second-phase particles. Because of the difference in alloy composition and casting method, it is necessary to find optimum homogenization cycle for each alloy and processing method.

Heat Treatments. In general, heat treatments determine the final mechanical properties of these materials. Solution heat treating, quenching and aging are basic heat treatments for 2xxx aluminum alloys. The proper selection of these heat treatments can achieve optimum combination of strength and ductility of the material.

A. Solution Heat Treating. The purpose of solution heat treatment is to put the maximum practical amount of hardening solutes such as copper, magnesium, silicon into solid solution in the aluminum matrix. Fig.2 is aluminum-copper equilibrium diagram. The composition of copper is less than 5.5% for most 2xxx aluminum alloys, such as 2014, 2018, 2024, 2025, 2124 and 2618 [1]. For these alloys, the solution temperature at which the maximum amount is soluble corresponds to a eutectic temperature. Overheating and

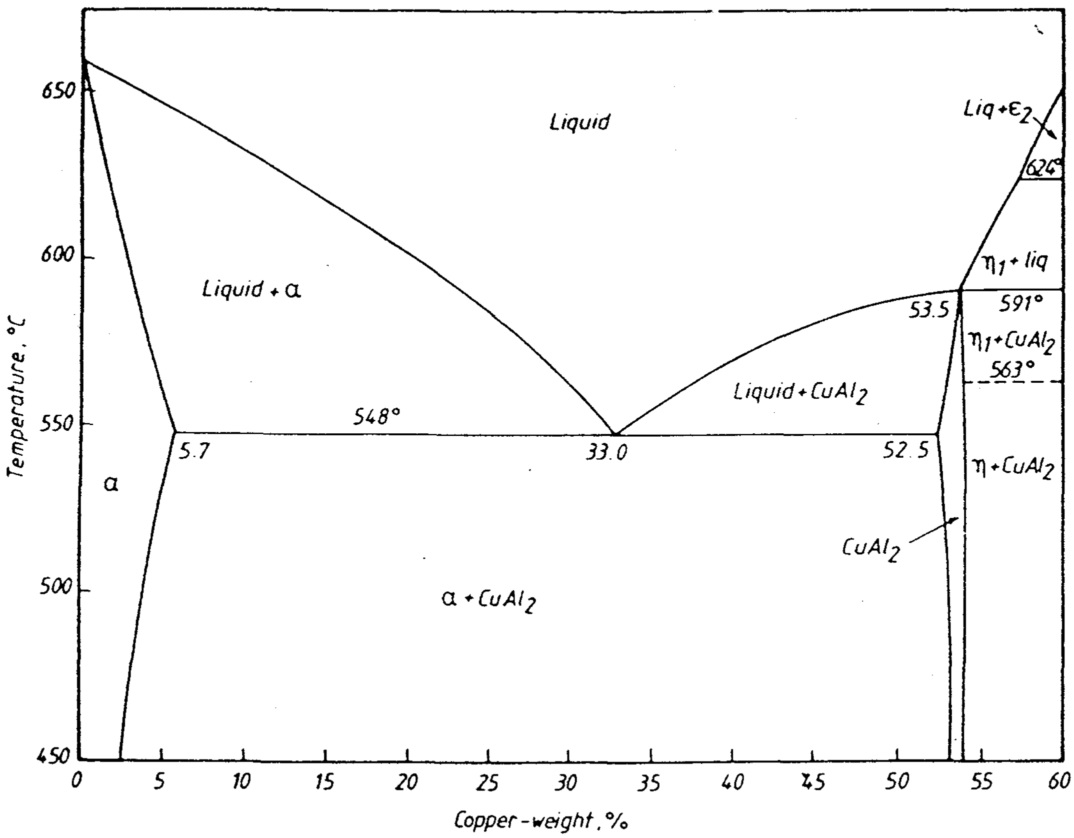
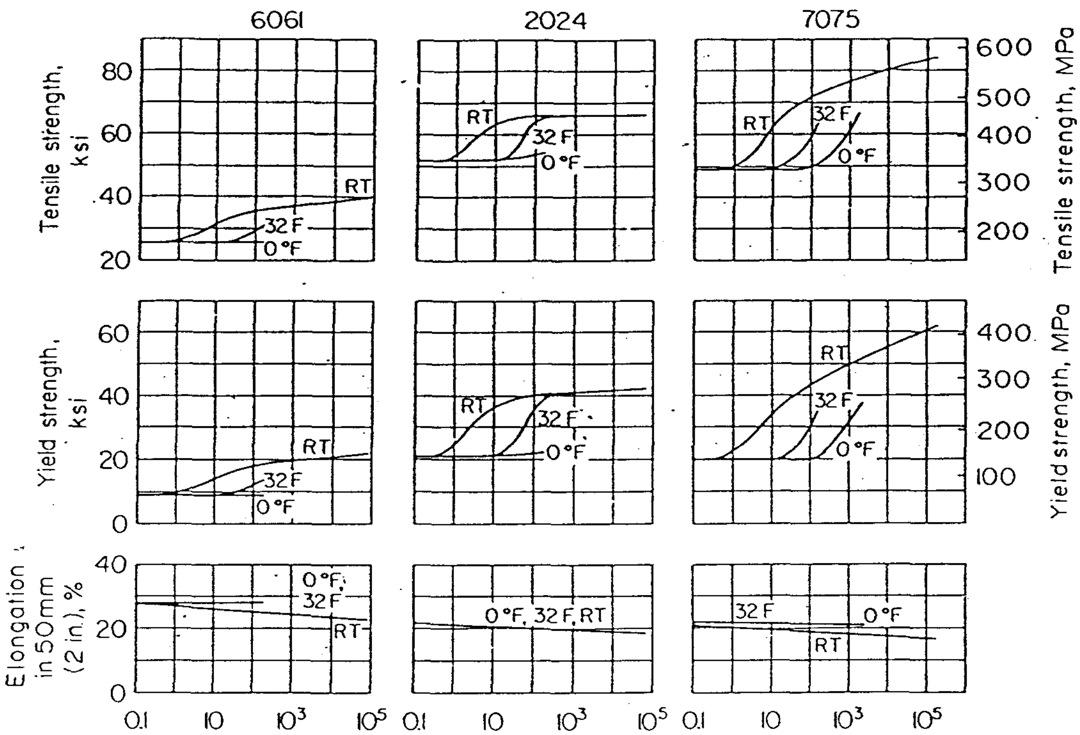


Fig.2 The Al-Cu Binary Phase Diagram [6]

partial melting must be avoided. The time required for solution heating depends on the type of product, alloy elements, casting or fabricating procedure used and thickness insofar as it influences the pre-existing microstructure. The main consideration is the coarseness of the microstructure and the diffusion distances required to bring about a satisfactory degree of homogeneity.

B. Quenching. Quenching is another important step in the sequence of heat treating operation. The purpose of quenching is to preserve the solid solution formed at the solution heat treating temperature by rapidly cooling to some lower temperature, usually near room temperature. Quenching not only retains solute atoms in solid solution, but also maintains a certain minimum number of vacancies that assist in promoting the low temperature diffusion required for precipitation. The rapid quenching rates improve the strength . The degree of distortion that occurs during quenching and the magnitude of residual stress that develops in the products tends to increase with the rate of cooling. In addition, the maximum attainable quench rate decreases as the thickness of the product increases. Because of these effects, much work [7,8,9,10] has been done over the years to understand and predict how quenching conditions and product form influence properties. Water is the most widely used and most effective quenching medium. Cooling rates can be reduced by increasing water temperature. 2xxx aluminum alloys that are to be artificially aged are often quenched in boiling water or oil to reduce distortion and residual stresses and maintain good resistance to corrosion at the same time [11].

C. Aging. The purpose of aging is to increase strength and resistance to corrosion by forming GP zones and precipitating second-phase particles from solid solution obtained from quenching. There are two types of aging for 2xxx aluminum alloys: natural aging and artificial aging. Most of the heat treatable alloys exhibit age hardening at room temperature after quenching, called natural aging. Microstructural changes accompanying natural aging are difficult to detect without transmission election microscipy because the



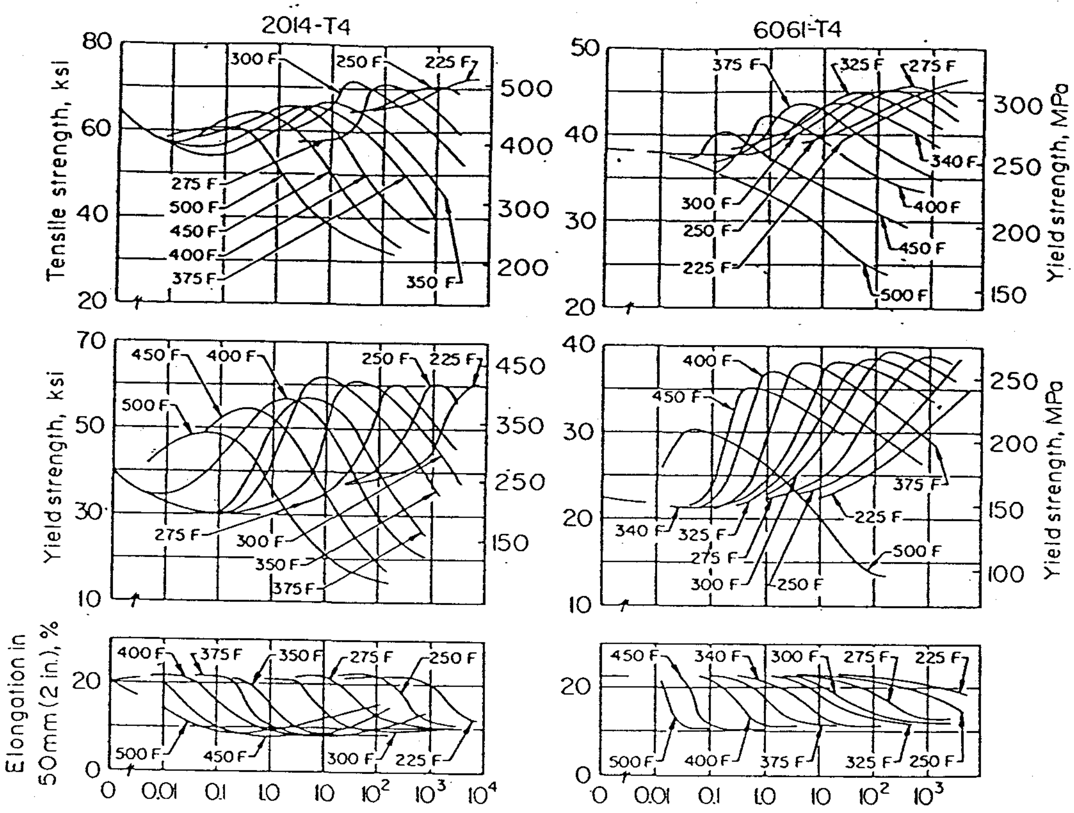
Aging Time (hrs)

Fig.3 The aging characteristics of aluminum alloys at room temperature, 0°C, and -18°C [5].

hardening effects are attributable solely to the formation of a zone structure within the solid solution. Most of the strengthening occurs within a few hours after quenching at room temperature for most 2xxx aluminum alloys [4]. The mechanical properties are usually essentially stable after four days. 2xxx alloys stored under refrigeration can retard aging. Fig.3 shows the aging characteristics of 2024 aluminum alloy at ambient and sub-ambient temperature.

The hardening observed at room temperature is attributed to localized concentrations of copper atoms forming Guinier-Preston zones, designated GP(1) [4,5]. These consist of two-dimensional copper rich regions of disk-like shape, oriented parallel to {100} planes. The diameter of the zones is estimated to be 3 to 5 nm and does not change with aging time at room temperature. The number, however, increases with time, until in the fully aged condition. The average distance between zones is about 100 nm. The electrical and thermal conductivities decrease with the progress of natural aging. Because a reduction of solid solution solute content normally increases electrical and thermal conductivities, the observed decreases are regarded as significant evidence that natural aging is a progress of zone formation, not true precipitation.

By reheating the quenched material to an elevated temperature, the solute content will be precipitated from solid solution, called artificial aging which greatly affects the mechanical properties of 2xxx aluminum alloys. A characteristic feature of artificial aging effects on tensile properties is that the increase in yield strength is more pronounced than the increase in tensile strength . Also, ductility and toughness decrease [4,5,11]. Overaging decreases both the tensile and yield strengths, but ductility generally is not recovered in proportion to the reduction in strength, so that the combinations of these properties developed by overaging are considered inferior to those prevalent in the underaged condition. Fig.4 illustrates the aging characteristics of 2014 aluminum alloy at elevated temperatures. Some of the important features illustrated are: a.) hardening can be retarded,



Aging Time (hrs)

Fig.4 The aging characteristics of two aluminum alloys at elevated temperatures [5].

or suppressed indefinitely, by lowing the temperature. b.) the rates of hardening and subsequent softening increase with increasing temperature. c.) over the temperature range in which a maximum strength can be observed, the level of the maximum generally decreases with increasing temperature. The mechanisms of artificial aging hardening differ from those of natural aging hardening. For aluminum-copper alloy at temperatures of 100°C and higher, the GP(1) zones disappear and are replaced by a structure designated GP(2), which although only a few atom layers in thickness, is considered to be three dimensional and to have an ordered atomic arrangement. The transition phase θ', having the same composition as the stable phase and exhibiting coherency with the solid solution lattice, forms after GP(2), but coexists with it over a range of time and temperature. The final stage in the sequence is the transformation of θ' into noncoherent equilibrium phase θ ($CuAl_2$). The structure sequence may be diagrammed:

$$SS \longrightarrow GP(1) \longrightarrow GP(2) \longrightarrow \theta' \longrightarrow \theta$$

In most 2xxx aluminum alloys the principle of age hardening is basically the same as in the pure aluminum-copper alloys, except that iron, manganese and silicon tend to reduce the rates of early stages of precipitation, so that the alloys show only limited hardening with natural aging. Small additions of magnesium accelerate and intensify natural aging hardening.

Age hardening mechanisms have been studied over many decades [8]. At relatively low temperatures and during initial periods of artificial aging at moderately elevated temperatures, the principal change is the redistribution of solute atoms within the solid-solution lattice to form GP zones. This local segregation of solute atoms produces a distortion of the lattice planes, both within the zones and extending for several atom layers into the matrix. With an increase in the number or density of zones, the degree of disturbance of the regularity and periodicity of the lattice increases. The strengthening effect of the zones results from the additional interference with the motion of dislocations when they cut the GP zones. This may be because of chemical strengthening (the production of a new particle-matrix interface) and the increase in stress required to move a dislocation through a region distorted by coherency stress. The progressive strength increase with natural aging time has been attributed to an increase in the size of the GP zones in some systems and to an increase in their number in others. As aging temperature or time are increased, transition precipitation happens. The strengthening effects of these transition structures are related to the impedance to dislocation motion provided by the presence of lattice strains and precipitate particles. Strength continues to increase as the size of these precipitates increases as long as the dislocations continue to cut the precipitates. Further progress of the precipitation reaction causes the structure of the precipitate to change from transition to equilibrium form. With loss of coherency strain, strengthening effects are caused by the stress required to cause dislocations to loop around rather than cut the precipitates. Strength progressively decreases with growth of equilibrium phase particles and an increase in interparticle spacing.

Recent studies have attempted to improve hot workability and mechanical properties of AA2024 aluminum alloy . In a previous study [12,13], the elevated-temperature torsional behavior of aluminum alloy AA2024 [1] was determined at relatively high strain rates. These tests were performed subsequent to both homogenization and precipitation treatments at various temperatures. The purpose of the work was to determine thermal pretreatments that lead to optimal extrudability which was presumed to occur where the elevated-temperature(torsional) stress versus strain behavior indicated lower peak stresses and higher ductility, the former implying lower break-through pressures. This would increase the limiting extrusion ratio of a given press and increase the permissible extrusion velocity [13-17].

These investigators [12,13] performed torsion tests at 310 and 370°C at a strain-rate, ε, of 1 s^{-1} on AA2024 alloy, homogenized between 420 and 520°C, and either water quenched or air cooled and subsequently aged at 290 or 350°C. The stress versus strain curves of the alloy in the quenched condition exhibited relatively large peak stresses,σ_p, and work softening with relatively low strain to failure. The peak and work softening was reduced and the ductility improved as either the aging temperature or the deformation temperature was increased; the behavior became closer to that of pure Al in which a high level of dynamic recovery provided excellent workability [16]. The homogenization temperature had a large effect on σ_p but little effect on the ductility for deformation at 310°C, whereas at 370°C there was little effect on σ_p but a large effect on ductility which was much higher. These results suggest that suitable homogenization and precipitation treatments can lower the elevated-temperature flow stress, notably decreasing the initial high peak stress. They concluded that the decrease in flow stress and increase in the ductility subsequent to precipitation treatments arise from the amounts of solute and dynamic precipitation [of, presumably,$CuAl_2$ and $(Cu,Mg)Al_2$ [13]]. Such a reduction in peak stress and increase in ductility would seem to be beneficial for direct extrusion of the alloy. In a review [18,19], McQueen showed that similar effects to the above had been found in other age-hardenable Al alloys.

The present study intended to verify the previous investigation [12,13] and expand the experimental scope. Basically, the intent of this study was to also determine the effect of different precipitation treatments on the hot workability of aluminum alloy AA2024. Instead of torsion tests, the present study emphasized elevated-temperature tensile tests which may be important in some extrusion processes. Additional precipitation temperatures were also considered. This study also considered the effect of the heating rate to the homogenization temperature, the time at homogenization temperature, the rate of cooling from the homogenization temperature, and the time at the precipitation temperature, to find their effects on the T3 and T8(ambient-temperature) tensile properties of AA2024. Furthermore, the "laboratory" conditions that predict favorable extrudability were commercially tested by directly extruding AA2024 alloy, and determining whether adequate ambient-temperature properties are evident subsequent to extrusion. Therefore, this work may be considered as a logical extension of the earlier work mentioned above.

2014 alloy, consisting of the same alloying elements as those of 2024, but less amount of silicon and magnesium, is a main commercial alloy of 2xxx aluminum alloys. The standard homogenization treatment, solution treatment and aging treatment are recommended by the Aluminum Association, Inc[1]. Homogenization treatment and subsequent heat-treatment greatly affect the mechanical properties of this alloy [4,5]. In order to improve the T3 properties of this alloy further, another intension of this study was to determine the effect of different homogenization cycles (including the homogenization temperature, the time maintained at the homogenization temperature, and the heat-up and cool-down rates) on the T3 properties of 2014 alloy.

2618 aluminum alloy, which was developed from Al-Cu-Ni-Mg alloy called "Y" alloy first created in Britain during the period of 1917 to 1919, has been widely used as a piston material since that time. 2618 aluminum alloy consists of most of the alloy elements in 2xxx alloys and can be extruded and mechanically worked hot in various ways. Virtues of 2618 alloy are its strength at high temperatures, its elastic properties, and good resistance to atmospheric and marine corrosion. The hardening which occurs on aging of 2618 is partly due to the precipitation of Mg_2Si, $Al_4CuMg_5Si_4$, and other complex constituent second-phase particles [6,20-21] and not of $CuAl_2$ as with 2024 since the copper becomes associated with the nickel to form a ternary intermetallic compound. The absence of the $CuAl_2$ constituent gives rise to the superior strength of 2618 alloy at high temperatures.

Recent Japanese work [22] discovered that increased copper concentration can significantly increase (15%) the elevated temperature tensile properties of the alloy discribed here as 2618 (Cu-rich). An additional objective of this research was, then, to determine the reliability of the Japanese work and determine whether changes in the homogenization temperature, the heating rate to, and the cooling rate from the homogenization temperature can improve the ambient-temperature (and presumably elevated temperature) T6 properties of 2618 and 2618 (Cu-rich). Furthermore, investigations on the effect of various T6 temperatures and times at temperature on the T6 properties for 2618 and 2618 (Cu-rich) were investigated. The composition of Cu in the 2618 (Cu-rich) was actually about 0.3wt% above the composition limits [1].

EXPERIMENTAL PROCEDURES

Chemical Composition and Ingot Size

This study utilized AA2024, 2014 and 2618 aluminum alloys. The aluminum alloys were provided in the form of direct chill cast ingots (where the solidification rates are 2 to 10 times faster than the traditional rates) of 83 mm and 178 mm diameter for AA2024, 83 mm diameter for 2014, and 98 mm for 2618 and 2618 (Cu-rich) alloys. The compositions of the ingots of AA2024, 2014, 2618 and 2618 (Cu-rich) aluminum alloys tested in this study are listed in Table 1.

Equipment

Tensile tests were performed on an Instron 4505 screw driven tensile machine with computerized data acquisition. The accuracy of most of the mechanical tests was within $\pm 0.5\%$. The homogenization treatments were performed in an air velocity controlled furnace with Partlow controller. The temperature were controlled within $\pm 4^{o}C$ of the set temperature. The solution treatments were performed in a case furnace with an accuracy within $\pm 2^{o}C$ of the set temperature. Elevated-temperature tests were performed utilizing a three-zone furnace with Eurotherm controllers and power supplies.

Table 1.

Table 1(a). Specification for the chemical composition of AA2024 alloy used in this study

Limit	Si	Fe	Cu	Mn	Mg	Cr	Zn	Ti	V	Be
Min wt%			4.15	.50	1.30					.001
Max wt%	.15	.30	4.35	.70	1.50	.05	.18	.05	.01	.005

Table 1(b). Specification for the chemical composition of 2014 alloy used in this study

Limit	Si	Fe	Cu	Mn	Mg	Cr	Zn	Ti	V	Be
Min wt%	.50		3.90	.40	.20					
Max wt%	1.2	.70	5.00	1.20	.80	.10	.25	.15	.05	.05

Table 1(c). Specification for the chemical composition of 2618 alloy used in this study

Limit	Si	Fe	Cu	Mn	Mg	Cr	Ni	Zn	Ti	B	Ca
Min wt%	.10	.90	1.90		1.30		.90		.04		
Max wt%	.25	1.30	2.70	.000	1.80	.05	1.2	.10	.10	.05	.05

Table 1(d). The chemical composition of 2618 (Cu-rich) alloy used in this study

Limit	Si	Fe	Cu	Mn	Mg	Cr	Ni	Zn	Ti	B	Ca
wt%	.10	1.36	3.01	.21	1.58	.05	1.30	.02	.075	.05	.05

Experimental Procedures

The tensile test specimen geometries varied and the typical gage dimensions for ingot characterization were 5.1 mm diameter and 25.4 mm length. Specimens from extrusions had 2.8 mm diameter and 20 mm length or 11.4 mm gage length. Specimens were evaluated from random positions within the ingots and extrusions. It was determined that the mechanical properties were independent of the positions. Heat treatment were T3 and T8 for AA2024 aluminum alloy, T3 for 2014 aluminum alloy and T6 for 2618 aluminum alloy. The T3 treatment consists of a solution treatment followed by a water quenching, refrigeration 1-3 hours with a subsequent plastic "stretch" of 1 to 2.5%, followed by an ambient temperature age. The "stretch" increases the density of dislocations that act as heterogeneous nucleation sites and increases the strength over the undeformed and aged (T4) alloy. Specimens were kept in the freezing compartment of a refrigerator to suppress precipitation. The T8 treatment was similar to T3 treatment except the aging treatment was an artificial-age instead of an ambient-temperature-age and the refrigeration time was less than 1/4 hr. The T6 treatment for 2618 consists of a solution treatment followed by a boiling water quench, 1 hour refrigeration, followed by an artificial age. For solution treatment, a 10 min heat-up was required to achieve the solution temperature (493°C for 2024 , either 493°C or 502°C for 2014 and 529°C for 2618 and 2618 (Cu-rich) aluminum alloys), once specimens were inserted into the furnace. Generally, tensile specimens were maintained at temperature for 1 hour. The ambient temperature age varied form 2.5 to 19 days for T3 treatment of AA2024 alloy and was 4 days. The aging temperature and time-at-temperature of T8 treatment for 2024 was 190°C and 12 hrs, respectively. The artificial aging of T6 treatment of 2618 alloy consisted of three aging temperatures(166°C, 182°C and 199°C) and various aging times up to 24 hrs.

The ductility was measured as engineering strain to failure (El) equal to $\Delta L/L_0$ where L_0 is the initial length and, for high-temperature tests, also as a reduction in area(RA) equal to $(A_0 - A_f)/A_0$ where A_0 is the initial area and A_f is the final area. The yield and ultimate tensile stress were reported as engineering stress. The yield stress was based on a 0.002 plastic strain offset. Strain rates varied between 0.67×10^{-3} to 3.0×10^{-3} s^{-1}.

RESULTS AND DISCUSSION

I. Study of AA2024 Aluminum Alloy

A. Elevated-Temperature Tests of Precipitation Treated Alloy

A set of tensile experiments were performed to compare with the Belgian work [12,13]. Specimens were extracted from 83 mm diameter homogenized ingots. Homogenized specimens received precipitation treatments at 270, 330, 350, and 370°C for 1, 6 or 12 hrs. This resulted in precipitation of $CuAl_2$ and $(Cu,Mg)Al_2$, based on the microscopy performed by the earlier work. This was followed by elevated-temperature testing at 350°C (a common extrusion temperature). From Table 2, Fig.5 and Fig.6, it appears, based on the 350°C tests on specimens that were not precipitation treated, that prolonged precipitation treatments decrease the tensile strength and increase the ductility. More specifically, the table and figure indicate the precipitation treatments between 330°C and 350°C for 12 hours reduce the yield stress and ultimate tensile strength up to 14% and increases the tensile ductility by a factor of up to 1.2. These tensile results are qualitatively consistent with the earlier work [12,13] that was performed by torsion testing.

Some scatter in the data may arise from the procedure of heating to the testing temperature. This heating was done as quickly as possible (30 min for our furnace) to "simulate" (rapid) induction heating that is frequently utilized in commercial extrusion processes. As a consequence, some non-reproducible temperature gradients existed along the specimen axes (e.g. 20°C variation). The resulting error in stress measurement is about $\pm$3MPa.

Table 2.

The elevated temperature mechanical properties of "precipitation-treated" alloy specimens from homogenized 83 mm dia. ingot.

Test Temp. (°C)	Precip. Temp. (°C)	Precip. Time (hrs)	σ_y (MPa)	UTS (MPa)	El (%)	RA (%)
350	270	12	69.6	84.1	29	86
350	330	1	75.8	89.6	33.3	89
350	330	6	79.6	91.7	36.3	88
350	330	12	66.2	82.0	35.3	79
350	350	1	76.5	89.6	35.4	89
350	350	6	69.6	84.1	36.6	88
350	350	12	62.1	79.3	33.4	88
350	0	--	76.5	89.6	30.3	84

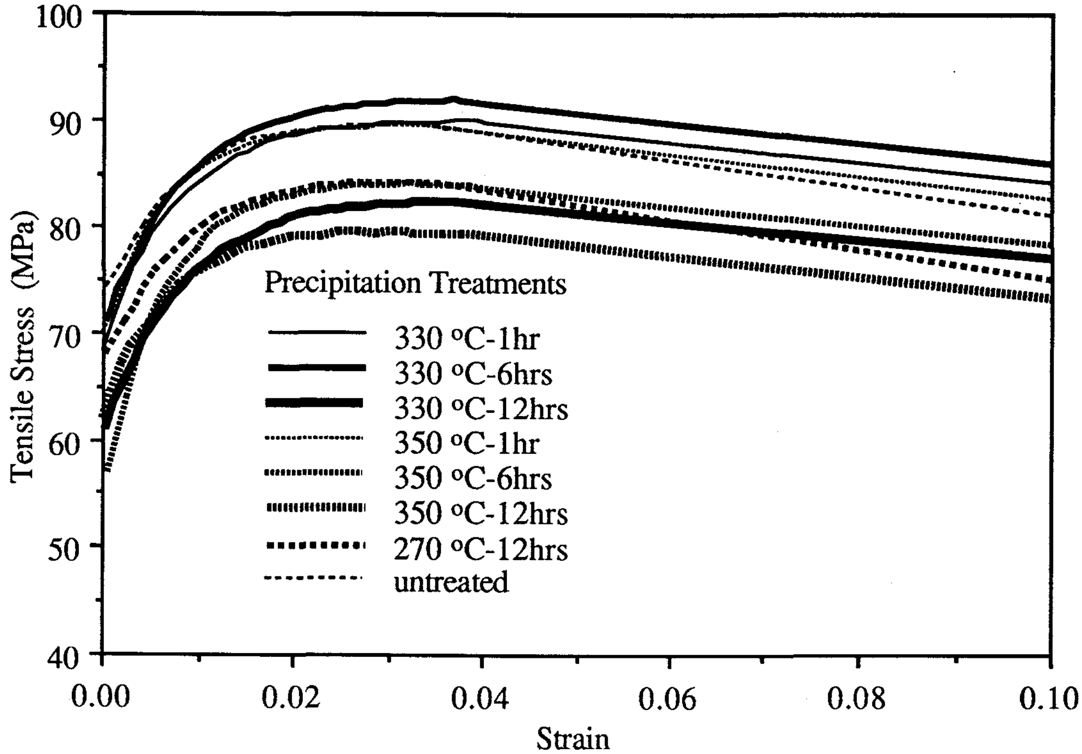


Fig.5 The elevated temperature tensile stress vs. strain curves of AA2024 alloy specimens from homogenized 83 mm dia. ingot ($\dot{\varepsilon}=10^{-3}\ s^{-1}$), subsequent to various precipitation treatments.

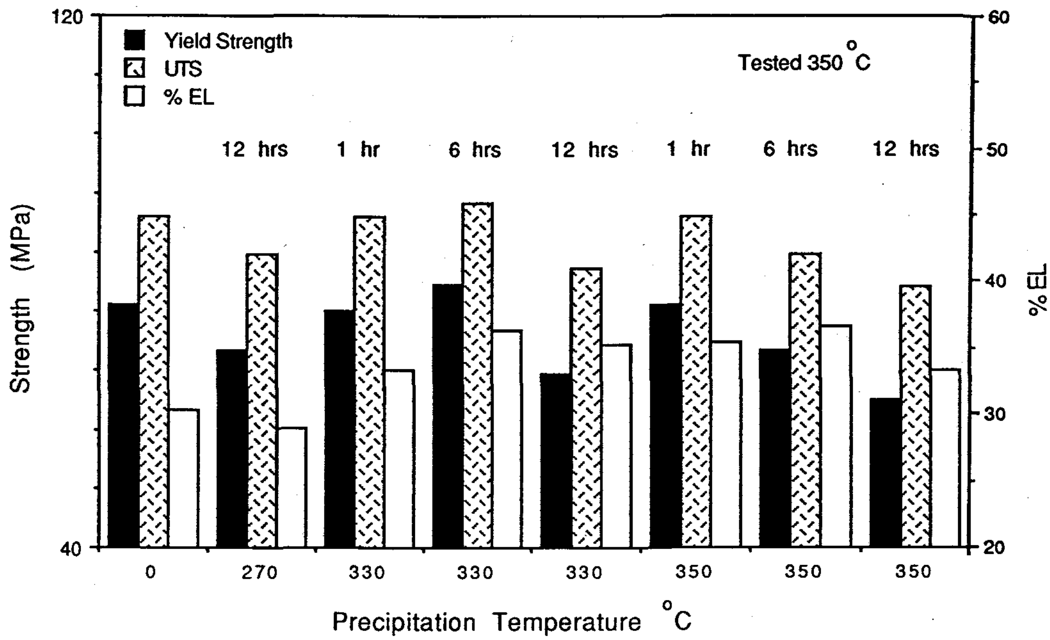


Fig.6 The relationship between precipitation temperatures and strength and elongation of AA2024 alloy specimens from homogenization 83 mm dia. ingot tested at 350 °C.

B. Elevated-Temperature Extrusions of the Precipitation Treated Alloys

Although the precipitation treatments appear to increase workability (specifically extrudability) based on simple mechanical tests, a more important test is extrusion testing. In the next series of tests, 178 mm diameter ingots were extruded and air cooled into "L"-shaped beams of 5.1 mm thickness, and 63.5 mm lengths (R=38.4). Ingots were extruded (billet temperature at 316-343°C after 10-15 min induction heating) after various commercially relevant thermal processes, some of which were believed to cause precipitation. The treatments included 1) a fan-cool (FC) with a cooling rate of 315-370°C/hr that followed a 6-hr homogenization (482-499°C with 8-10 hrs linear heating to the homogenization temperature); 2) oven-cool (OC) after an identical homogenization as with 1) with an additional 1-hr linear reheating to 504°C with a 12-hr linear cooling rate; and 3) an oven-step-cool, where subsequent to the homogenization in 2) the ingot is cooled to 349°C (a "precipitation" temperature) in 4 hrs and maintained for 6 hrs before an 8-hr linear cooling to ambient temperature.

As expected, it appeared that the oven-step -cool ingots could be most easily extruded in terms of the maximum possible extrusion speed of the ingot with acceptable surface quality. With this criteria, it is expected that softer, perhaps more ductile alloy, would be most extrudable. The extrusions of fan-cooled ingots showed tearing at the surface at 7.3 m/s while oven-step-cool extrusion did not give evidence of tearing until about 8.8 m/s. Extrusions of oven-cooled ingots (probably some precipitation) did not show tearing at 8.5 m/s but showed tearing at 9.1 m/s. This is in agreement with improved extrudability of billets with prior precipitation treatment observed in 2024 [17] and in Al-Mg-Si alloys [16,23]. Table 3 and Fig.7 show the tensile test results of T3 and T8 specimens extracted parallel to the axis of the extrusion at random locations. The difference

Table 3.

T3 and T8 properties of AA2024 specimens from 178 mm dia. ingot of AA2024 alloy, extruded after various thermal treatments. ("L"-shape extrusion)

Cooling Method	Treatment	σ_y (MPa)	UTS (MPa)	El (%)
FC	T3	359	445	13.6
OC	T3	340	420	15.1
OSC	T3	341	418	15.5
FC	T8	474	495	6.3
OC	T8	473	483	4.4
OSC	T8	472	482	4.6

FC - furnace cool

OC - oven cool

OSC - oven-step-cool

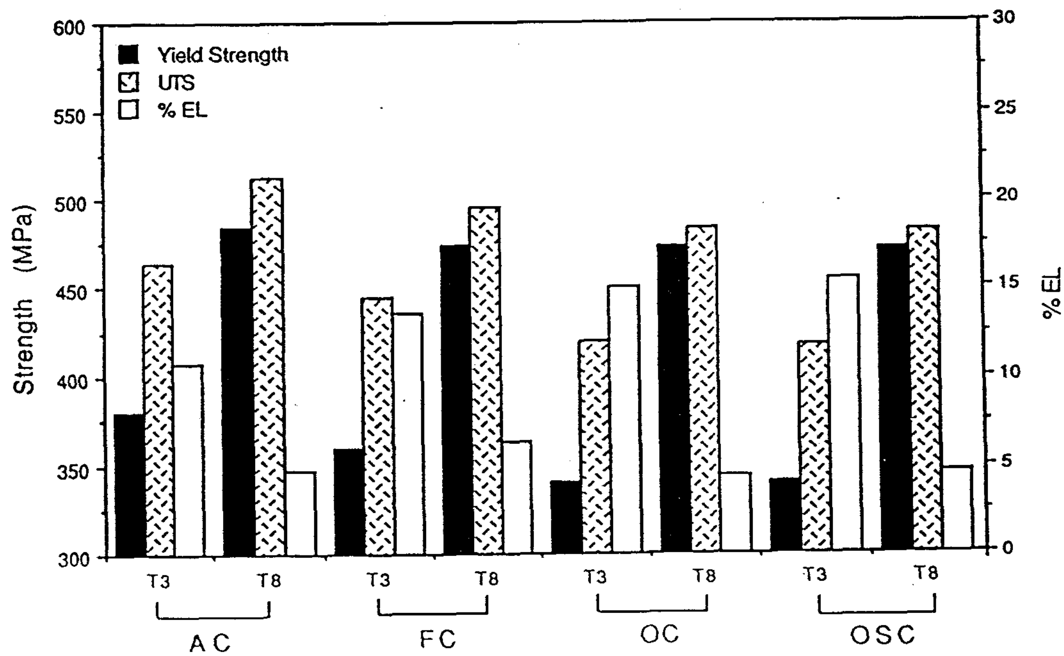


Fig.7 T3 and T8 properties of AA2024 specimens from 178 mm dia. ingot, extruded after various thermal treatments.

in T3 tensile properties, aged from 2.5 to 19 days, was not substantial; only 4 MPa (roughly 1%) for the yield stress; 6 MPa for the ultimate tensile strength (also 1%); and 1.5% El (factor of 1.1) [24], consistent with other work [5]. The mechanical properties listed are an average of five tests. Table 3 shows that the "final" T3 and T8 properties are essentially identical for the three "pretreatments". Therefore, there is encouraging evidence that the oven-step-cool pretreatment can increase the extrudability and yet not compromise the critical T3 and T8 mechanical properties (at least for this starting ingot dimension, extrusion temperature, and final configuration).

Experiments analogous to the above were performed on precipitation-treated ingots of a smaller diameter (83 mm) and extruded into a simple cross-sectional configuration of 7.9 mm by 32.5 mm with "round" corners (R=21.1). The oven-cooled (OC) ingots were linearly heated to the homogenization temperature within 12 hrs. The oven-step-cooled (OSC) ingots were also maintained at 504°C for 4 hrs (after a 12 hr heating) and linearly cooled to 349°C (the same "precipitation" temperature as with the "L"-shaped extrusions) within 1.5 hrs, maintained at 349°C for 4 hrs, and cooled to 93°C in 4 hrs before removal from the furnace. The insert temperature was between 370 and 425°C (about 50°C higher than the previous ("L") extrusion temperature). The induction heating of the extrusion required about 15 min. It was found that the OSC billets could be processed at a greater extrusion speed without surface defects than conventionally homogenized 2024 (although the quantitative rates were not reported). Table 4 and Fig.8 show averages of three ambient-temperature tensile tests for each cooling cycle on extrusions and homogenized as-cast ingots which provide a reference. First, the T3 (2.5% stretch, 19 days) and T8 mechanical properties of homogenized, unextruded ("as cast") ingots were determined. The homogenization included a 9-hr linear heating to the homogenization temperature of 504°C, maintained at temperature for 6 hrs, and cooling at 150°C/hr to ambient temperature. The

Table 4.

T3 and T8 properties of specimens extracted from 83 mm dia. AA2024 ingots, extruded subsequent to various thermal treatments.

Cooling Method	Treatment	σ_y (MPa)	UTS (MPa)	El (%)	RT Age Time (Days)
OC	T3	431	543	13.7	12.5
OC	T3	432	541	13.0	19
OSC	T3	409	508	13.8	12.5
OSC	T3	408	513	15.5	19
AC	T3	380	464	10.8	19
OC	T8	492	523	9.9	--
OC	T8	498	529	9.8	--
OSC	T8	490	518	8.8	--
OSC	T8	496	525	9.7	--
AC	T8	484	512	4.7	--

OC - oven cool

OSC - oven-step-cool

AC - as-cast

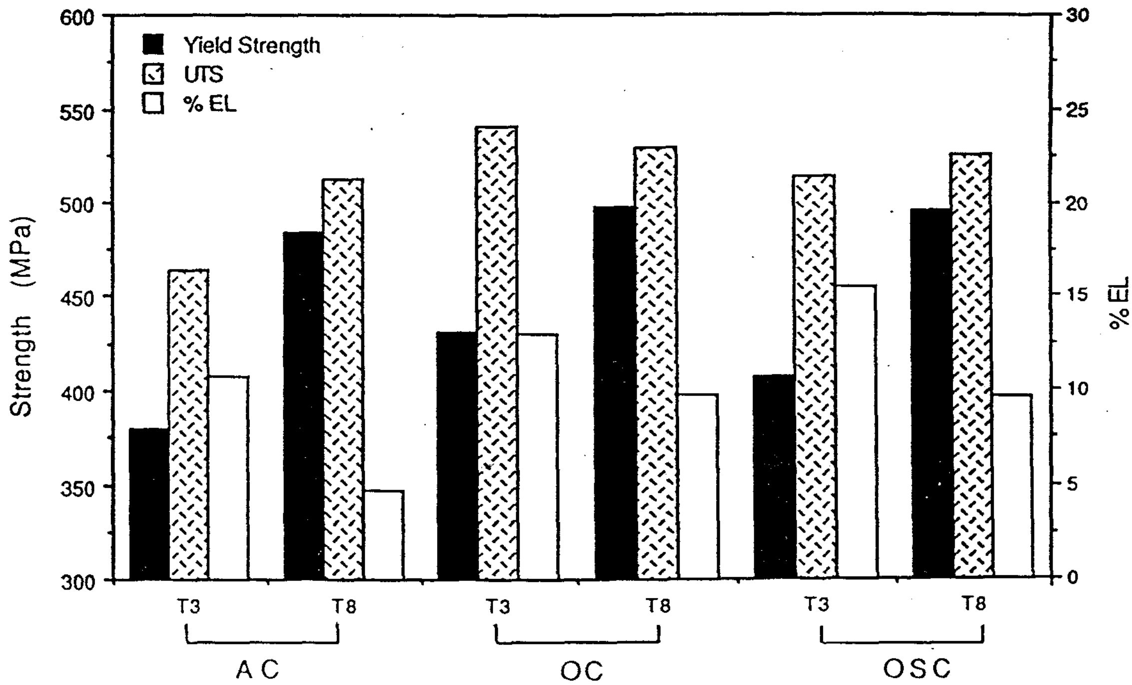


Fig.8 T3 and T8 properties of specimens extruded from 83 mm dia. AA 2024 ingots, extruded subsequent to various thermal treatments.

variation of the test results was very small, ±1%. Since these ingots were not extruded, the tests do not reflect production properties. The strength values (average of 6 tests) are slightly higher than the typical AA2024 T3 and T8 values and the ductility is slightly lower [1,25]. The T3 and T8 properties of the OC and OSC extrusions were well above average. We note that the previous "L" extrusions have lower strength than both the baseline (non-extruded) material and extrusions from 83 mm ingots, indicating the variability of properties resulting from different ingot sizes and extrusion processes.

The correlation between high ductility in tension and in extrusion seems reasonable. Previous work on 2024 [15,17] does not confirm this; material cooled from the homogenization temperature to the billet insert temperature had higher strength and breakout pressure, lower ductility, but higher maximum extrusion velocity than furnace cooled material. Part of the problem lie in the lack of precise knowledge of the distribution and concentration of constituent particles and of precipitate materials in both the present and previous research. Another part lies in the extrusion process in which there is an average rise of 50 to 100°C in average temperature in the deformation zone, it being larger for higher billet stresses (lower insert temperature, higher alloying or particle content) and larger extrusion ratio and ram speed [14,15]. However, there is an additional temperature increase from the average to that of the periphery as it passes near the die land. The problem has been resolved in a detailed study of Al-Mg-Si alloys by Reiso [26,27]; in slowly cooled billets large Mg_2Si precipitates form on constituent particles and do not dissolve on preheating or in the deformation zone but form a eutectic with the constituent particles which melt at the die land, causing tearing. In billets rapidly cooled to an intermediate temperature, precipitates are smaller and more homogeneous with the result that they redissolve in preheating and deformation so that no eutectic is formed and the tearing temperature is close to the true solidus of the alloy at a higher ram speed. This theory may be applicable to the present 2024 alloys but the distribution and nature of constituent particles and location or size of precipitates in different cooling cycles have not been determined. All research is agreed that proper control of the precipitation produces lower flow stresses at 300-400°C and lower extrusion breakout pressures [12-13,15-17, 19, 23, 26-29].

C. Effect of the Duration of the Precipitation Treatment on the Ambient-Temperature Mechanical Properties

Another part of this study consisted of determining the precipitation time on the T3 mechanical properties. The homogenization cycle consisted of linear heating from ambient temperature to 504°C within about 8 hrs. The specimens (of about 12.7 mm x12.7 mm x152 mm that were cut from a 83 mm dia. ingot prior to homogenization) were homogenized for 6 hrs at 504°C. The samples were then step-cooled by a 3-hr linear cooling to 350°C, the precipitation temperature, and maintained for various times ranging up to 72 hrs. Subsequently the samples were removed from the furnace and air cooled. The T3 properties were then assessed after an age of 12.5 days.

The results illustrated in Fig.5 show that for all precipitation treatment times up to 72 hrs (limit considered commercially practical), higher yield stress, and UTS are associated with higher ductility. Relatively favorable T3 properties are observed with very prolonged precipitation times (e.g., greater than 48 hrs), as well as at times of about 20 hrs, and times up to about 6 hrs. Lower (but not unacceptable) T3 properties are observed for precipitation times between 8 and 12 hrs. The microstructural (e.g., precipitation) features that are associated with these changes were examined by optical metallography but the observations were not conclusive. It must, of course, be emphasized that the details of Fig.6 may change at the earlier times (e.g., less than serval hours) due to changes in sample dimensions, since longer times are required for heating the interior of thicker

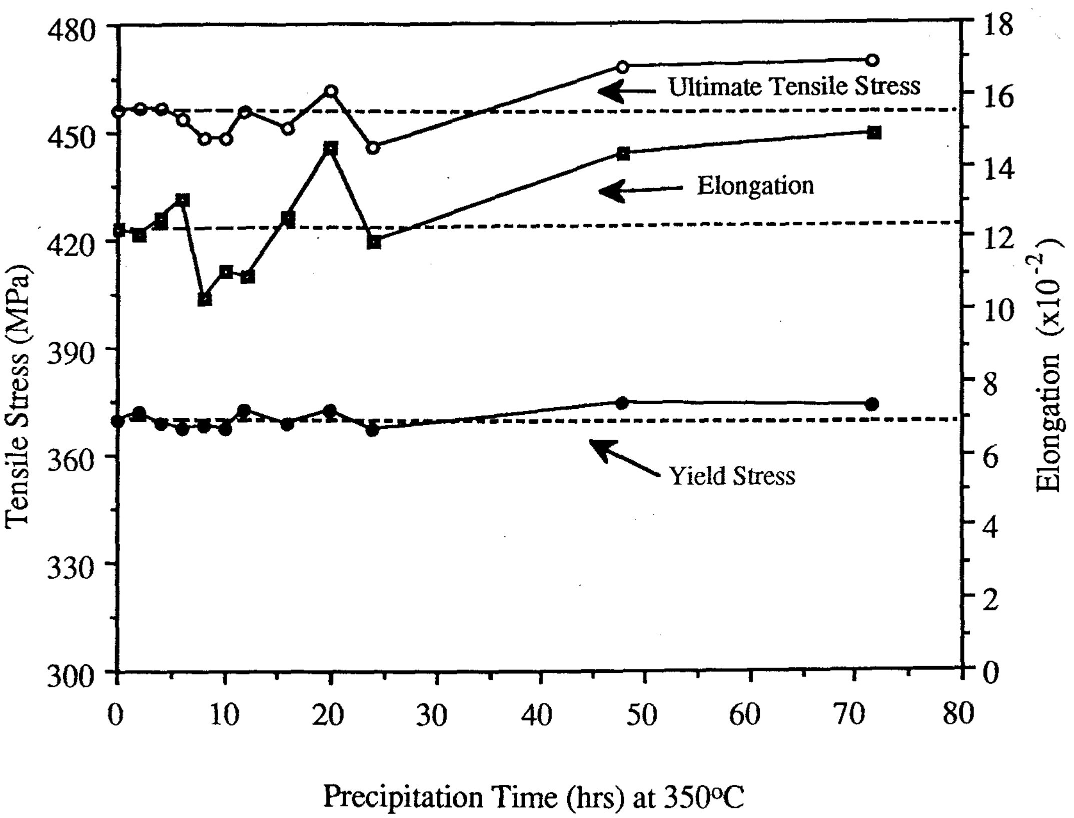


Fig.9 The effect of precipitation time at 350°C on the T3 tensile stress and elongation of AA2024 aluminum alloy.

specimens to the desired temperature. Also, the results may be different due to subsequent mechanical processing, as shown with the earlier extrusion tests.

Just as the data of Fig.5 suggest that up to 6 hrs and over 40 hr precipitation treatments will optimize T3 properties, Table 5 suggests that these same times (at least up to 12 hrs) at 350°C increase workability. These two observations are consistent with the extrusion experiments (4-6 hr. "precipitation" treatments), where favorable extrudability and T3 properties were observed. This indicates that the large particles formed during prolonged aging dissolve during the standard solution treatment. However, prolonged precipitation may impede solution during passage through the extrusion press in press heat treatment of Al-Mg-Si (6xxx series) alloys [16, 23, 26-29].

The long-range rise in strength and ductility with time cannot be related to precipitation directly because of the subsequent solution treatment. It can, perhaps, be related to continued homogenization. Paterson and Sheppard [30] have noted progress of dissolutions at 400°C; large precipitate and constituent particles become smaller and regions of eutectic became fewer up to 24 hrs. The Belgian work showed that ductility rose as a result of 460°C annealing as well as a result of 500°C, but that in 24 hours the hot strength and segregated particle density [$CuAl_2$, $(FeMn)_3Si_2Al_6$] at 460°C decreased only half as much as at 500°C [13]. For 8 hours homogenization, the ductility at 350°C rose as the temperature increased from 440°C to 520°C [12]. While the diffusion rate is lower, there is still be a tendency for useful alloying elements to redistribute from constituent particles or eutectic into the uniform distribution of large precipitates; those atoms would then be available for hardening upon heat treatment. It is also possible that at 300°C there would be some coalescence of the dispersed particles which would decrease the amount of heterogeneous precipitation thus enhancing homogeneous formation and improved hardening.

D. The Effects of Various Homogenization Cycles on the Ambient-Temperature Mechanical Properties

Analogously, we determined the effect of various homogenization times (at 504°C) as well as heating rates (to the homogenization temperature) and various cooling rates to ambient temperature on the T3 (2.5% stretch, 4 days) tensile properties. The selected times for nearly linear heating to the homogenization temperature were 4, 8, and 12 hrs; the duration of the homogenization times were 6,12, and 24 hrs; and the times for linear cooling to ambient temperature were 4 (-120°C/hr), 8 (-60°C/hr), and 12 (-40°C/hr) hrs.

The results are listed in Table 5; each value is an average of 3 tests. This table indicates that the T3 properties are not dramatically sensitive to our implemented changes in heating-rate, homogenization or "soak" time, or to cool-down times for cast ingots. Table 6 and fig.10 provide some insight into some subtle independent effects of the heat-up times (rates), the homogenization times, and the cool-down times (rates). The average σ_y, UTS, and % El (9 average values as derived from Table 5) are listed for each heating time, saturation time, and cooling time. The data suggests that the 8-hr heating followed by the maximum homogenization time of 24 hrs followed by a relatively rapid cooling within 4 hrs be expected to produce superior T3 properties. (However, 8 or 12 hours cool could be better for improved extrudability (Sec.B) with negligible change in properties.) In fact, Table 5 reports that this combination produced T3 properties of 374 MPa yield stress, 640 MPa ultimate tensile stress, and 13.3% elongation. These values are somewhat better than the average values reported in Table 6. The explanation for these trends is not readily apparent.

Homogenization treatments to shrink and spheroidize constituent particles and to remove coring, non-equilibrium particles, and eutectics are usually more complete for longer times and higher temperature [31]. The observed behavior is in agreement with the

Table 5.

T3 tensile properties of 2024 alloy extracted from 83 mm dia. ingot heated to the homogenization temperature, 504°C. Solution temperature was 493°C.

	Heat-up Time (hrs)	Soak Time (hrs)	Cool-Down Time (hrs)	Yield Stress (MPa)	UTS (MPa)	El (%)
1	4	6	4	374	461	13.4
2	4	12	4	374	458	13.2
3	4	24	4	371	456	14.1
4	4	6	8	377	460	12.0
5	4	12	8	373	461	12.9
6	4	24	8	372	457	13.9
7	4	6	12	369	452	11.2
8	4	12	12	373	447	11.5
9	4	24	12	369	454	14.1
10	8	6	4	374	461	13.7
11	8	12	4	377	463	14.1
12	8	24	4	374	461	13.3
13	8	6	8	374	461	12.4
14	8	12	8	375	462	13.4
15	8	24	8	374	462	13.6
16	8	6	12	372	456	13.4
17	8	12	12	371	456	12.9
18	8	24	12	369	454	13.7
19	12	6	4	376	463	13.2
20	12	12	4	375	461	13.3
21	12	24	4	373	458	13.4
22	12	6	8	374	461	11.8
23	12	12	8	372	461	13.1
24	12	24	8	373	458	12.9
25	12	6	12	374	456	12.3
26	12	12	12	371	456	13.3
27	12	24	12	370	455	13.3

Table 6.

Effect of heat-up, soak, and cool-down times on the T3 mechanical properties of AA2024 alloy specimens.(from Table 5)

	σ_y (MPa)	UTS (MPa)	El (%)
Heat-Up Times (hrs)			
4	372	457	12.9
8	374	460	13.4
12	373	459	13.0
Time at the Homogenization Temperature (hrs)			
6	374	459	12.6
12	374	459	13.1
24	373	458	13.6
Cool-Down Time (hrs)			
4	374	460	13.5
8	374	460	12.9
12	371	455	12.9
Over-All Average	373	458	13.1

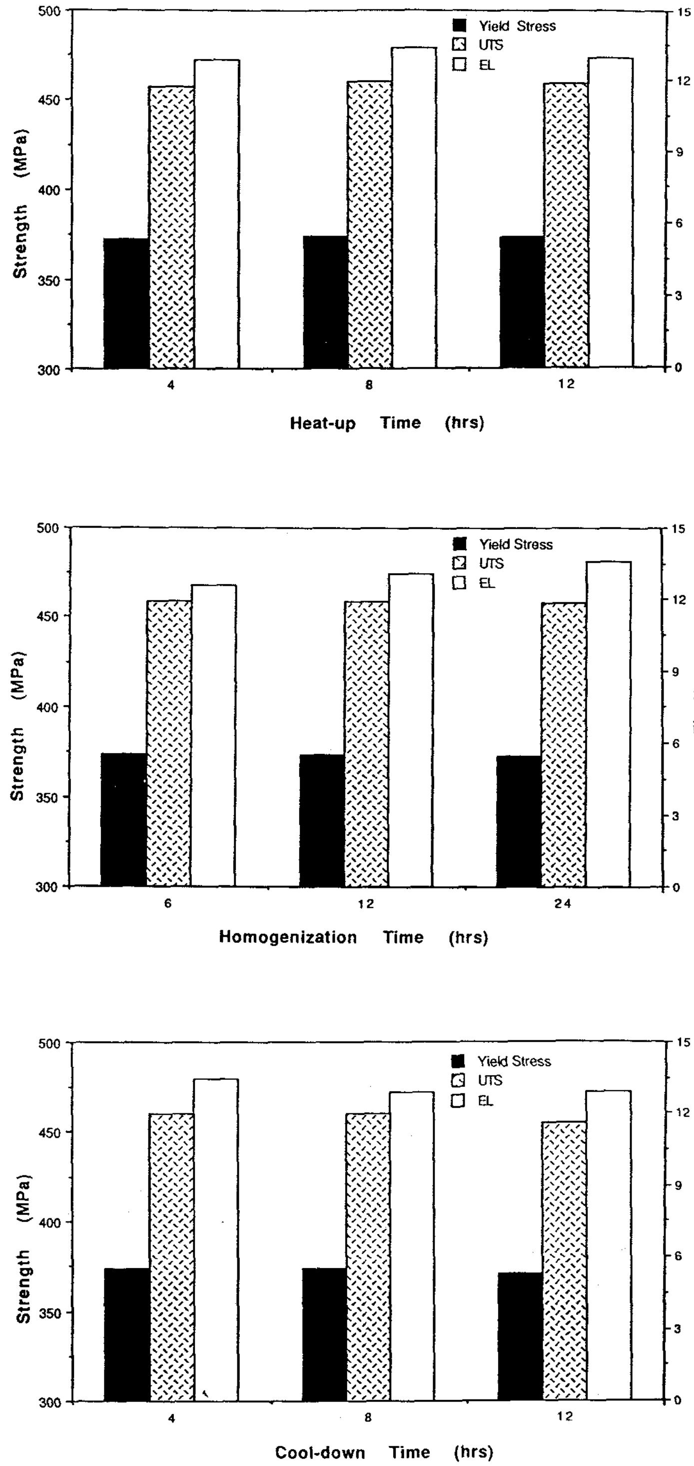


Fig.10 Effect of heat-up, soak and cool-down times on the T3 mechanical properties of AA 2024 alloy specimens.

Belgian work that the strength and the density of particles at the grain boundaries decrease as time rises from 8 to 24 hours [13] and the ductility at 350°C rises as the temperature increases from 440 to 520°C [12]. Homogenization treatment of 24 hrs have been shown to reduce the flow stress of 2014 by 20% [31]. The disappearance of the $CuAl_2$ eutectic and reduction of the Mg_2Si eutectic was observed. Constituent particles of $Cu_2Mg_8Si_6Al_5$ and $(FeMn)Al_6$ became smaller in size as redistribution of soluble elements took place (Mg_2Si particles diminished to 1.5 μm in 8 hrs and to 0.6 μm in 24 hours). In Al-Mg-Si alloys, Precht and Pickens [32] showed that 4 hr homogenization 550°C produced the maximum improvement in ductility. At 550°C, ductility improved up to 6 hrs and remained stationary from there to 10 hrs; however, the strength was hardly allied at all. It was also well known that faceted β(AlFeSi) gradually changes over long times to spheroidal α(AlFeSi) which is much less deleterious to ductility [31].

II. The Study of 2014 Aluminum Alloy

A. The Effect of Various Homogenization Cycles on the Ambient-Temperature Mechanical Properties

Analogous tests to determine the effect of various heating rates (to the homogenization temperature), homogenization times as well as various cooling rates to ambient temperature on T3 tensile properties were performed on AA2014 aluminum alloy. The specimens were extracted from 83 mm dia. ingots and the homogenization cycles were the same as those of AA2024 alloy studied above in Sec.D. The two solution treatment temperatures (493 and 502°C) were used in this study. The specimens were linearly heated to the homogenization temperature within 4, 8, 12 hrs; maintained at the homogenization temperature for 6, 12, 24 hrs; and then linearly cooled down to 93°C within 4, 8, 12 hrs followed by air cooling to ambient temperature. Only two heat-up times (4 and 12 hrs) and two cool-down times (4 and 12 hrs) were used for the specimens at the solution treatment temperature of 502°C.

Table 7 and Table 8 list the results of the specimens solution treated for 1 hr at 493°C and the recommended temperature of 502°C [1], respectively. Each value is an average of 3 tests. It is noted from Table 9 and 10 that the 9°C increase in solution temperature (from 493 to 502°C) under the same homogenization treatment not only increases the yield stress values by about 10.3 MPa (3% increase), the UTS by 17.2 MPa (4% increase), but also improves the elongation from 11.5% to 13.5%. The results indicate that different solution temperatures appear to affect the T3 tensile properties of 2014 aluminum alloy; higher solution temperature increasing both strength and ductility of the alloy.

The results suggest that the T3 tensile properties of 2014 alloy are, as with AA2024 alloy mentioned above, generally insensitive to changes in heating-rate, homogenization time, or to cooling-down rate for cast ingots with the same solution temperature. Table 9 and Table

Table 7.

T3 tensile properties of 2014 alloy extracted from 83 mm dia. ingot heated to the homogenization temperature, 504°C. Solution temperature was 493°C.

	Heat-up Time (hrs)	Soak Time (hrs)	Cool-Down Time (hrs)	Yield Stress (MPa)	UTS (MPa)	El (%)
1	4	6	4	341	441	10.6
2	4	12	4	338	439	11.6
3	4	24	4	334	434	11.4
4	4	6	8	337	435	11.0
5	4	12	8	338	438	10.7
6	4	24	8	336	435	11.5
7	4	6	12	339	439	11.4
8	4	12	12	338	440	12.6
9	4	24	12	332	431	11.3
10	8	6	4	341	441	11.2
11	8	12	4	339	439	11.7
12	8	24	4	334	434	11.4
13	8	6	8	341	440	10.8
14	8	12	8	337	438	11.4
15	8	24	8	334	434	11.8
16	8	6	12	337	439	11.7
17	8	12	12	333	435	11.8
18	8	24	12	332	431	11.0
19	12	6	4	341	443	11.8
20	12	12	4	338	438	11.5
21	12	24	4	335	434	11.8
22	12	6	8	341	443	12.1
23	12	12	8	341	445	13.1
24	12	24	8	335	434	11.5
25	12	6	12	339	442	12.6
26	12	12	12	335	436	12.2
27	12	24	12	330	434	12.7

Table 8.

T3 tensile properties of 2014 alloy extracted from 83 mm dia. ingot heated to the homogenization temperature, 504°C. Solution temperature was 502°C.

	Heat-up Time (hrs)	Soak Time (hrs)	Cool-Down Time (hrs)	Yield Stress (MPa)	UTS (MPa)	El (%)
28	12	12	8	348	455	14.0
29	8	6	4	350	460	13.3
30	4	6	12	350	459	13.3
31	4	6	4	347	452	11.5
32	4	12	4	350	455	13.5
33	4	24	4	343	450	13.3
34	12	6	4	352	462	13.6
35	12	12	4	350	459	13.9
36	12	24	4	345	454	14.0
37	4	12	12	348	454	13.6
38	4	24	12	342	447	13.7
39	12	6	12	349	456	13.2
40	12	12	12	347	454	13.9
41	12	24	12	343	448	14.1

Table 9.

Effect of heat-up, soak, and cool-down times on the T3 mechanical properties of 2014 alloy specimens (from Table 7). Solution temperature was 493°C.

	σ_y (MPa)	UTS (MPa)	El (%)
Heat-Up Times (hrs)			
4	337	434	11.3
8	336	436	11.4
12	337	439	12.1
Time at the Homogenization Temperature (hrs)			
6	339	440	11.5
12	337	439	11.8
24	334	434	11.6
Cool-Down Time (hrs)			
4	338	438	11.4
8	338	438	11.5
12	335	439	11.2
Over-All Average	337	437	11.5

Table 10.

Effect of heat-up, soak, and cool-down times on the T3 mechanical properties of 2014 alloy specimens (from Table 8). Solution temperature was 502°C.

	σ_y (MPa)	UTS (MPa)	El (%)
Heat-Up Times (hrs)			
4	349	453	13.2
12	347	456	13.8
Time at the Homogenization Temperature (hrs)			
6	350	457	12.9
12	349	456	13.7
24	343	450	13.8
Cool-Down Time (hrs)			
4	347	455	13.3
12	347	453	13.6
Over-All Average	347	454	13.5

10 show the average σ_y, UTS and % El (9 average of values as derived from Table 7 and 8, respectively) for each heating time, saturation time, and cooling time. The results suggests that for the standard solution annealing temperature, the maximum heating time and homogenization time of 12 hrs followed by a relatively rapid cooling may improve T3 properties. Compare to the results of AA2024 alloy, longer heat-up times (slower heat-up rates) and shorter homogenization times appear to provide superior T3 tensile properties of 2014 aluminum alloy. These results are approximately consistent with those of AA2024.

III. Study of 2618 Aluminum Alloy

A. The Effect of Various Homogenization Cycles on the Ambient-Temperature Mechanical Properties

Recent Japanese work [22] indicated that increased copper concentration of the standard AA2618 alloy can significantly increase (15%) the elevated temperature tensile properties of the alloy. An additional objective of this research was, then, to determine the reliability of this work and determine whether changes in the homogenization temperature, heating rate to, and cooling rate from, the homogenization temperature can improve the ambient-temperature (and presumably elevated temperature) T6 properties of standard 2618 and the "new" 2618 (Cu-rich). Furthermore, investigations on the effect of various T6 temperatures and times at temperatures on the T6 properties for 2618 and 2618 (Cu-rich) were also investigated. The composition of Cu in the 2618 (Cu-rich) was actually about 0.3wt% above the standard composition limits [1].

Three homogenization cycles were performed to determine the effect of homogenization cycles on the T6 tensile properties of 2618 and 2618 (Cu-rich) aluminum alloys. Homogenization Cycle 1 consisted of linear heating to 500°C within 4 hrs, maintaining at 500°C for 12 hrs, and linear cooling to 93°C within 8 hrs followed by air cooling to ambient temperature. Homogenization Cycle 2 was the same as the Homogenization Cycle 1 except the homogenization temperature was 541°C instead of 500°C. Homogenization Cycle 3 consisted of linear heating to 493°C within 4 hrs, maintaining at 493°C for 10 hrs, linear heating to 529°C in 1 hr, soaking at 529°C for 1 hr, and linear cooling to 93°C within 8 hrs [33]. The subsequent T6 heat treatment was the same for all specimens homogenized under the three homogenization cycles. The solution temperature and aging temperature were 529°C and 199°C, respectively. The aging time was 20 hrs plus 10 min heat-up.

Table 11 lists the tensile test results of T6 specimens of 2618 and 2618 (Cu-rich) aluminum alloys homogenized under three homogenization cycles. Each value is the average value from ten T6 specimens. From Table 11, it appears that homogenization cycles affect the T6 tensile properties of the alloys. Homogenization Cycle 2 not only increases the tensile strength, but also increases the ductility up to 10% compared with Homogenization Cycle 1 and Homogenization Cycle 3. Because Homogenization Cycle 2 had a higher homogenization temperature than the other homogenization cycles, it is possible to predict that the higher homogenization temperature (homogenization time is same) is associated with a better T6 tensile properties of 2618 and 2618 (Cu-rich) alloys.

We also note from Table 11 that different alloying elements and the amount of alloying elements affect the T6 tensile properties. 2618 (Cu-rich) alloy had a better strength (5% increase in yield stress) but a lower ductility (14% decreasing in elongation) compared to 2618 alloy in all of three homogenization cycles.

B. The Effect of Aging Treatment on the Ambient-Temperature Mechanical Properties

In order to study the effect of aging treatment on the T6 tensile properties of 2618 (Cu-rich) aluminum alloy, the specimens were treated at one of three different aging temperatures (166°C, 182°C and 199°C, the latter being the standard [1] temperature), removed from the furnace after 4, 8, 12, 16, 20, 24 hrs for each temperature and air cooled to ambient temperature. The Homogenization Cycle 2, linear heating to 541°C, maintaining at 541°C for 12 hrs, and linear cooling to 93°C followed by air cooling to ambient temperature, was used as the homogenization treatment. The samples were solution treated at 529°C for 1 hr, followed by a boiling water quench, and then maintained in refrigerator for not longer than 1hr before aging treatment.

The Fig.11 illustrates the T6 tensile test results of 2618 (Cu-rich) aluminum alloy under the three different aging temperatures and various aging times up to 24hrs.

Table 11.

Table 11(a). The T6 mechanical properties of 2618 and 2618 (Cu-rich) alloys under Homogenization Cycle 1. Solution temperature was 529°C.

Alloy	Treatment	σ_y (MPa)	UTS (MPa)	El (%)
2618	T6	358	441	7.45
2618 (Cu-rich)	T6	381	449	6.44

Table 11(b). The T6 mechanical properties of 2618 and 2618 (Cu-rich) alloys under Homogenization Cycle 2. Solution temperature was 529°C.

Alloy	Treatment	σ_y (MPa)	UTS (MPa)	El (%)
2618	T6	361	446	8.38
2618 (Cu-rich)	T6	378	450	7.75

Table 11(c). The T6 mechanical properties of 2618 and 2618 (Cu-rich) alloys under Homogenization Cycle 3. Solution temperature was 529°C.

Alloy	Treatment	σ_y (MPa)	UTS (MPa)	El (%)
2618	T6	355	441	7.50
2618 (Cu-rich)	T6	377	448	6.74

Each value is the average value of three tensile specimens. Fig.11 shows that yield stress and ultimate stress increases but ductility decreases as aging time increases at the aging temperatures of 182°C and 199°C. The strength and ductility change very slowly at the low aging temperature of 166°C. At constant aging time, the higher aging temperature is associated with a higher strength and a lower ductility. As aging time increases, yield stress increases more greatly than ultimate tensile stress increases (e.g., the yield stress increases up to 15% at 182°C and 35% at 199°C, but the ultimate tensile stress increases 2.5% at 182°C and 8% at 199°C as aging time increases from 4 to 24 hrs.) This result is consistent with the studies of most 2xxx alloys [1,4,5,11], indicating that a characteristic feature of elevated-temperature aging effects on tensile properties is that the increase in yield strength is more pronounced than the increase in ultimate tensile strength.

It is also noted from Fig.11 that the strength-increase "rate" decreases between 16 hrs and 24 hrs at 199°C indicating that the maximum aging hardening is being approached at this temperature but overaging does not occur. The maximum aging hardening level is not obtained at 166°C and 182°C even up to 24 hrs. Therefore, the higher temperatures and longer times are required to get optimum aging hardening of 2618 (Cu-rich) alloy. It should be mentioned that the slight difference of values in Table 11 and Fig.11 is perhaps caused by the selection of different parts of ingot or some scatter.

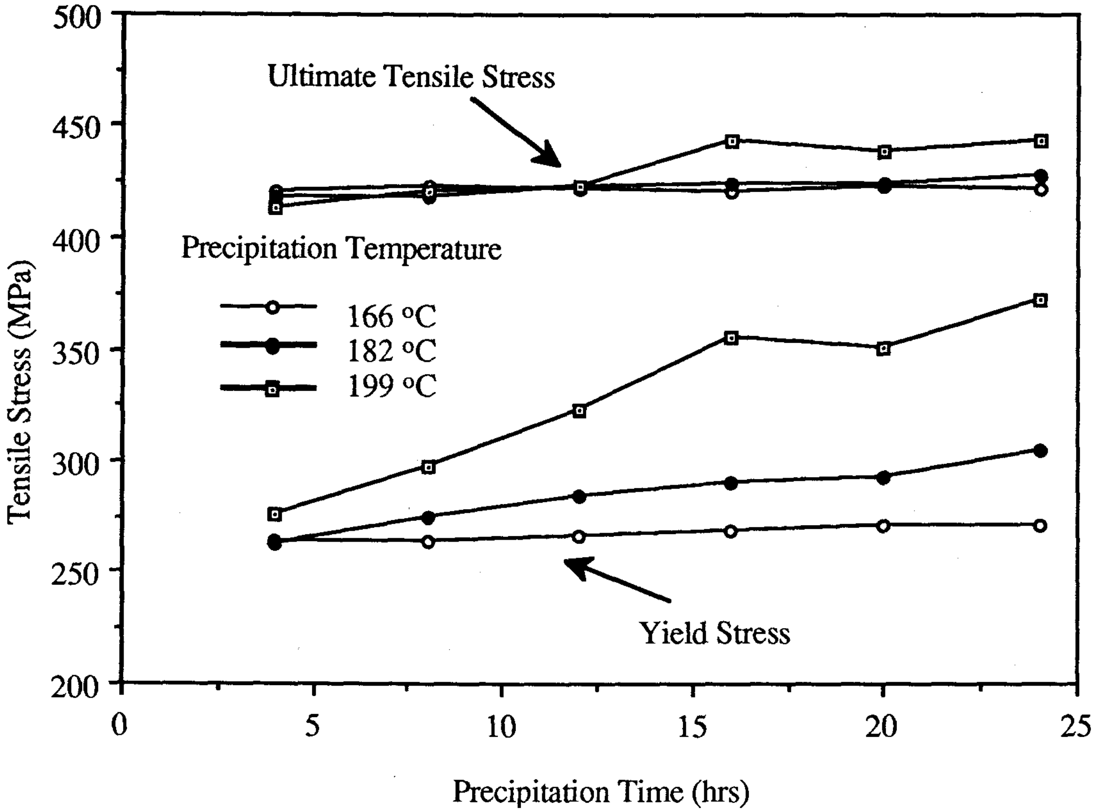


Fig.11(a) The effect of aging treatment on the T6 tensile strength of 2618 (Cu-rich) alloy.

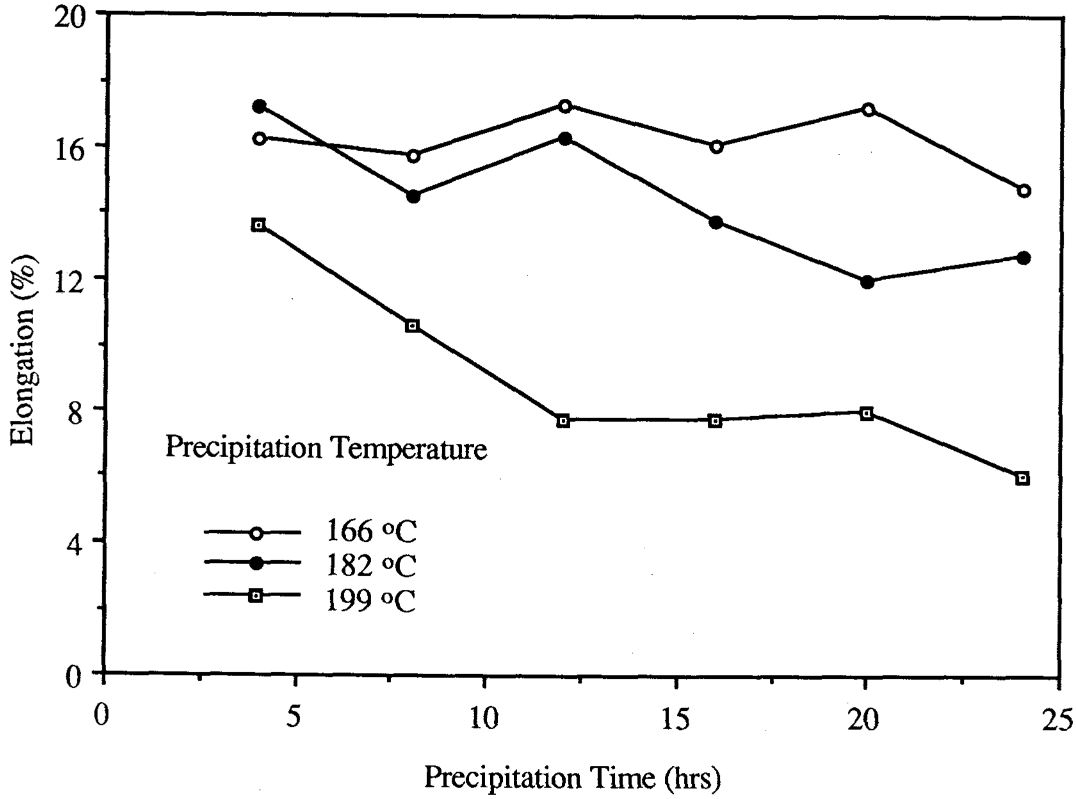


Fig.11(b) The effect of aging treatment on the T6 elongation of 2618 (Cu-rich) alloy.

CONCLUSIONS

1. Elevated-temperature tensile and extrusion tests indicate that the hot workability of AA 2024 aluminum alloy appears to improve with some elevated-temperature precipitation treatments.
2. The precipitation treatments before extrusion do not appear to degrade the ambient-temperature T3 and T8 tensile properties of AA2024 alloy.
3. The time at the precipitation temperature appears to affect the T3 and T8 tensile properties alloy in unextruded ingots of AA2024 aluminum alloy with longer times (e.g., > 40 hrs) providing both relatively high strength and ductility.
4. The heating and cooling rates to and from the homogenization temperature and the time at the homogenization temperature do not dramatically affect the T3 tensile properties of AA2024 and 2014 aluminum alloys. However, longer soak times and more rapid cooling rates may slightly improve the properties of these alloys.
5. Higer homogenization temperatures appear to improve the T6 tensile properties of 2618 and 2618 (Cu-rich) aluminum alloys, increasing both strength and ductility.
6. Adding manganese and slightly increasing the amount of copper appears to increase the T6 strength (specifically yield stress), but slightly decrease the ductility of 2618 alloy.
7. Aging temperature lower than the standard value does not improve the T6 properties over the standard aging time.

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