THERMOMECHANICAL PROCESSING
OF ALUMINUM ALLOYS

by

ALDO MARIO RETI

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Signature of Author
Department of Metallurgy

Signature of Professor
in Charge of Research

Signature of Chairman of
Departmental Committee on
Graduate Students

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ABSTRACT

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ALDO MARIO RETI

Submitted to the Department of Metallurgy on May 14, 1970 in partial fulfillment of the requirements for the degree of Doctor of Science.

A study is described of the structure, solution kinetics and mechanical properties of unidirectionally solidified Al-4.5 percent Cu and 7075 ingots after various thermomechanical treatments. The thermomechanical treatments consisted of deforming sections of a cast ingot various amounts by rolling at selected temperatures, followed by solution treatment of several lengths of time.

It was found that plastic deformation by rolling altered the cast structure in the two alloys in different ways. In Al-4.5 percent Cu alloy, only relatively few non-equilibrium second phase particles were broken into smaller ones. Most particles were unbroken and were little deformed. Interparticle spacing (a measure of diffusion distance) was affected only slightly by working. In 7075 alloy, the second phase was broken down into small particles arranged in bands. Band spacings (measures of diffusion distances) were significantly reduced by working.

Solution kinetics of Al-4.5 percent Cu alloy were influenced only slightly or not at all by working at the various selected temperatures. Solution kinetics of 7075 alloy were markedly enhanced by working. This difference between the two alloys was a result of the different structures produced by working. Solution kinetics of 7075 alloy were in quantitative agreement with a simple analysis presented here; solution kinetics of Al-4.5 percent Cu alloy were in qualitative agreement with an analysis based on a theory of diffusion-limited precipitation.

The mechanical properties of cast and wrought Al-4.5 percent Cu and 7075 alloys were found to improve by decreasing the amount of second phase. Higher temperatures of rolling yielded larger values of ductility in material containing large amounts of second phase; ultimate and yield strengths were affected very little by the rolling temperature.

Thesis Supervisor: Merton C. Flemings
Title: ABEX Professor of Metallurgy and Materials Science
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Scanning electron fractograph of 7075 alloy rolled 4/1 at room temperature and solutionized 19 hours at 860°F.
I. INTRODUCTION AND LITERATURE SURVEY

(a) General Introduction

During the last few years, many investigations have been
devoted to develop high strength aluminum alloys. But little success
was obtained in developing alloys with high strengths together with
large values of ductilities. For example, extruded powder compacts
with yield strengths as high as 117,500 psi and 4 percent elongation
were obtained by Roberts.¹ Haar² reported tensile strengths up to
125,000 psi, but with lower ductility. Antes et al.³ demonstrated
the influence of second phase particles on wrought aluminum alloys
of the 7000 series and showed that a dramatic improvement in ductili-
ty is possible by eliminating these second phases. Bower, Singh and
Flemings⁴ combined alloy development programs with good solidifica-
tion and metal working techniques; four different alloys tested showed
yield strengths in excess of 90,000 psi with elongations in excess
of 5 percent.

At the Italian Light Metals Institute, Di Russo⁵ system-
atically prepared complex Al-Zn-Mg-Cu alloys with high contents of
Zn+Mg+Cu. Combining proper aging treatments with additions of anti-
recrystallizers (Cr and Mn), he was able to obtain wrought products
with exceptionally high mechanical properties. More recently,
Conserva et al.⁶ investigated new thermomechanical treatments in
which they plastically deformed an alloy in stages intermediate to
the aging stages. Electron microscopy revealed the presence of a
stable dislocation structure in the alloy, which interacted favorably
with subsequent plastic deformation, resulting in high strengths together with large values of ductility.

Singh and Flemings\(^7,8\) explained the effect of thermo-mechanical processing on solution kinetics and mechanical properties of 7075 aluminum alloy. But very little is known about the direct effect of hot working operations on kinetics as well as properties; the present investigation is aimed to gain some knowledge on this matter.

Because of the already existing diverse information on high purity Al-4.5 weight percent Cu and 7075 alloys, which contain non-equilibrium second phases formed during solidification, these alloys were chosen as subjects for this study.

(b) Recovery and Recrystallization Processes

Hot working is defined as deformation under conditions of temperature and strain rate such that recovery processes take place simultaneously with the deformation.\(^9\) Because of this, hot working occurs at an essentially constant flow stress; and because the flow decreases with increasing temperature, the energy required for deformation is generally much less for hot working than for cold working.

Hot working is usually referred in practice to deformation above about 0.6 Tm, where Tm is the melting temperature in degrees Kelvin. It is employed extensively in industry, partly because large reductions can be effected at low stresses with little or no intermediate annealings.

Jonas, Sellars and Tegart\(^10\) recently wrote a complete
review on strength and structure under hot working conditions. They pointed out that large strains can be achieved with little or no strain hardening, indicating that dynamic softening processes can operate sufficiently rapidly to balance the strain hardening processes.

McQueen formulated a theory of hot working from experimental observations on aluminum, and discussed its applicability to various metals. The same author studied the hot working characteristics of aluminum, nickel, copper, and brass (same crystal structure) and showed that they recrystallize at different rates, according to variations in their stacking fault energies. Stüwe briefly reviewed the mechanisms of recrystallization (migration of large angle grain boundaries) and of recovery (movement of dislocations out of their slip planes) at high temperatures. He concluded that there is some experimental evidence for recrystallization during very slow deformation, for instance, in creep. But experiments on copper which came closer to actual hot working conditions seemed to indicate, however, that at higher deformation rates, recovery processes determined the strength of the material.

Jonas, McQueen and Wong distinguished two kinds of recovery processes: (1) static recovery, which refers to polygonization driven by internal stresses in conjunction with thermal activation, and (2) dynamic recovery which refers to polygonization during deformation, driven by the external stresses in addition to the internal stresses and thermal activation. These authors made hot extrusions of aluminum; strength and microhardness measurements indicated dynamic recovery produced considerably more softening than static
recovery.

Morris\textsuperscript{15} also reviews thermal softening mechanisms. He points out that during an annealing or thermal softening process, a deformed metal may undergo many changes before it is reduced to the completely softened state. These changes are normally described by the terms recovery, polygonization, primary recrystallization and secondary recrystallization. He points out that in alloys of aluminum in which second phase particles are present, the recrystallized grains first observed frequently form adjacent to the second phase particles in the early stages of primary recrystallization and then develop outward from these centers. This may be due to differences in the worked state of the material adjacent to the particles, since it has been observed that slip lines sometimes are terminated by the second phase particles.

Leguet, Whitwham and Herenguel\textsuperscript{16} compared the different amounts of worked structure obtained as a function of temperature of deformation, while avoiding recovery which would have taken place after deformation. These experiments were made by rolling and drawing aluminum, copper and brass. Recrystallization which took place during quenching was found to be important for OFHC copper, but was insignificant for aluminum because a stable structure was obtained by recovery during hot deformation.

Varley\textsuperscript{17} made a study of recovery and recrystallization of aluminum within the temperature range of 200 to 325°C and reported a value for the activation energy of the process. The isothermal annealing curves (ultimate tensile strength plotted versus logarithm of
time) were continuous curves exhibiting no discontinuity corresponding to the onset of recrystallization.

The high temperature deformation of aluminum-magnesium alloys was studied by Cotner and Tegart.\textsuperscript{18} They applied various deformation schedules to simulate hot working operations, and concluded that both flow stress during deformation and ductility are related to competition between dynamic recovery and static recrystallization.

Wusatowski\textsuperscript{19} found that the recrystallization of stainless steel occurring immediately after hot working (under isothermal conditions) depends on the amount of reduction, strain rate, and grain size. He derived an empirical formula and a nomogram to calculate the recrystallization rate of austenitic steel for all reductions and temperatures usually employed and for strain rates normally applied. Hot rolling of mild steels were investigated by Ohmori,\textsuperscript{20} who found that the microstructures of the hot rolled steels are closely related to the recovery and recrystallization of austenite. It was reported in the same work that the yield stress depended on the grain size and dislocation density of ferrite.

Morris et al. made some research on deformation of aluminum alloys, and in particular of Al-4 percent Cu alloy. Howard and Morris\textsuperscript{21} showed that a dynamic structural instability due to strain affects the deformation texture and work hardening characteristics of Al-4 percent Cu. They hypothesized that this structural instability is composed of at least two parts: (1) that due to a redistribution of solute, (2) that due to a change in distribution of line defects. The same authors found\textsuperscript{22} that the disposition of a solute is a structural
factor that significantly controls the recrystallization kinetics in the Al-4 percent Cu alloy; the interaction of precipitate with dislocations being a paramount structural factor in this regard. The annealing characteristics of some aluminum alloys have been studied by Morris and Walls, who stated that a significant variation in recrystallization kinetics only occurs when there is a variation in the intensity of pinning of dislocations by precipitates.

(c) Deformation of Two-Phase Structures During Working

In practical situations, the materials usually are not pure metals or solid solutions but are complex, containing second phases, either in massive or dispersed form. These phases can either be metallic (formed by precipitation or phase separation), or non-metallic (inclusions arising from impure base metal or from contamination during melting). As Jonas et al. point out, on subsequent hot working such second phases can affect strength and ductility. Further, during hot working, massive phases can be broken up or precipitation accelerated so that strength alters during hot working and the resultant structure is different from the initial structure. These structural changes can be controlled by controlling the hot working operation.

In 1937, Unckel made a study of the deformation of two-phase alloys by cold rolling. He chose systems in which the inclusions were: (1) harder than the matrix and plastic (beta crystals in alpha brass); (2) harder than the matrix but brittle (most intermetallic
compounds in aluminum alloys; (3) softer than the matrix and plastic (lead in brass); and (4) softer than the matrix but brittle (graphite in grey iron). He concluded that because of the different relative flow of inclusion and matrix, additional plastic flow had taken place in the vicinity of inclusions.

Pickering\(^{25}\) carried out an investigation of the deformation of various types of non-metallic inclusions during hot rolling of Fe-\(\text{O}_2\)-Si alloys. He showed that at high temperatures siliceous inclusions deform and elongate prior to fracturing but at lower temperatures they fracture with little or no deformation. He concluded that fracture of the inclusions is necessary to disperse the larger detrimental inclusions into smaller, less harmful particles.

Ductile fracture of copper containing cuprous oxide inclusions was studied.\(^{26}\) An attempt was made to correlate the breakdown of the matrix-inclusion interface with void initiation and propagation. The void shape and position relative to the tensile axis suggested an interface breakdown mechanism of initiation.

Edelson and Baldwin\(^ {27}\) found that in two-phase alloys consisting of particles and voids dispersed in a copper matrix, ductility depends upon volume fraction of the dispersed phase. This dependence was true for various size particles of the same composition, for similar size particles of different composition, for particles of different shapes, and for voids. In good agreement with these findings were Gurland and Plateau,\(^ {28}\) who assumed three different stages in ductile fracture: (1) formation of cracks at inclusions, (2) growth of the cracks, (3) internal stricture leading
to failure. The major part of the elongation took place during the second stage.

An excellent contribution on the mechanics of fracture from inclusions has been made by McClintock.\textsuperscript{29} But not in agreement with McClintock is Thomason,\textsuperscript{30} who presented a theory for ductile fracture by internal necking cavities.

Mogford\textsuperscript{31} compiled a very good review on the deformation and fracture of two-phase materials. The review is restricted to the materials relevant to physical metallurgy, i.e., metals, metal compounds, and ionic solids. In general, it is concerned with systems in which the second phase component is in the form of dispersed particles.

Singh and Flemings\textsuperscript{8} investigated the influence of thermo-mechanical treatments on the mechanical properties and fracture behavior of high purity 7075 aluminum alloy. They found that undissolved alloy second phase was the most important single variable influencing mechanical properties. Microcracks and holes were invariably found to originate in or around second phase particles.

Iwai et al.\textsuperscript{32} hot rolled steel specimens, prepared by powder metallurgy technique, therefore uniformly containing a certain amount of oxide inclusions. They found that those complex oxide inclusions with melting points nearly the same as the rolling temperature were deformed plastically, while other oxide inclusions with higher melting points showed brittle fracture.

Malkiewicz and Rudnik\textsuperscript{33} carried out an examination of the deformation on non-metallic inclusions during hot rolling of steel.
Assuming that the inclusions deformed to an ellipsoid shape during rolling, they defined an index of relative plasticity as

\[ \gamma = \frac{2/3}{\log \frac{A}{h}} \]

where:

\[ \gamma = \text{index of relative plasticity} \]
\[ \lambda = \frac{b}{a} \]
\[ h = \frac{A_0}{A_1} \]
\[ 2a = \text{inclusion major axes parallel to line joining roll axes} \]
\[ 2b = \text{inclusion major axes parallel to rolling direction} \]
\[ A_0 = \text{cross-sectional area of the steel before rolling} \]
\[ A_1 = \text{cross-sectional area of the steel after rolling} \]

This method to measure the relative plasticity was advantageously used by Maunder and Charles,\(^3\) who studied the behavior during soaking and hot rolling of inclusions in a 0.2 percent carbon steel ingot. The rolling was carried out under very close control at temperatures between 700°C and 1350°C. Because small size specimens were used for rolling, they were inserted into the front of larger pieces of 1/2 inch bar in order to prevent excessive heat loss (Figure 1). The larger bar acted as a large thermal capacity to conserve heat and prevent chilling in the rolls. The majority of inclusions were either (1) glassy manganese aluminosilicates, which were fluid at the highest rolling temperatures and exhibited brittle or plastic characteristics at lower temperatures, or (2) manganese sulphide inclusions which were plastic and deformed rather less than the steel under all rolling
conditions but exhibited maximum relative plasticity at low rolling temperatures. Under certain conditions, both sulfide and silicates could be induced to join together.

In the same work, they compare their results with those of Chao, who found that the sulfide was always softer than the steels; whereas Maunder and Charles found that sulfides deformed less than steel. Chao's relative hardness curves showed a sharp "kink" at the steel transformation temperature, but a much smaller kink was found by Maunder and Charles from relative plasticity measurements. An explanation for this difference in hardness and rolling behavior was that the inclusion deformed by the frictional forces between the steel and the inclusion, and a smooth-surfaced inclusion like the sulfide would have been expected to deform less than straightforward hardness measurements indicated.

Charles and Uchiyama confirmed the behavior of silicate inclusions during hot rolling, previously reported by Maunder and Charles. They used a model system in which silicate was introduced from a slag bath into small holes in pure iron specimens. These specimens were rolled at temperatures between 700°C and 1250°C under close control, enabling inclusion deformation behavior to be isolated into the various types described by Maunder and Charles.

Still interested in the relationship between hardness and deformation, Chao and Van Vlack placed small manganese sulfide inclusions with known crystallographic orientations inside powder compacts of low-carbon steel. They found that the relative deformation of the inclusion and metal was closely dependent upon the relative
hardness of the two phases.

(d) Mechanical Properties of the CuAl₂ Compound

In order to investigate the hot-deformation behavior of Al-4.5 percent Cu, and 7075 alloy it is advisable to learn as much as possible about the compounds always present in the alloy after solidification, a major one being CuAl₂. Petty¹⁷ found an inflection temperature, T₁, in the hardness temperature curves of several intermetallic compounds. Most intermetallic compounds were very brittle, and hardness indentations performed below T₁ were accompanied by prolific cracking. Above T₁ the indentations were not usually cracked in spite of the fact that more deformation occurred. It is therefore suggested that two different modes of deformation operate above and below T₁. T₁ for CuAl₂ was reported as 300°C.

Savitsky³⁹ has shown for several intermetallic systems, including CuAl₂, that the temperature of maximum strength in tension and compression is usually 50 to 80 percent of the absolute melting point of the system (melting point of CuAl₂ is 864°C), and that at higher temperatures there is an exponential decrease in strength. Crossman, Yue and Vidoz⁴⁰ hot rolled Al-CuAl₂ eutectic composites and pointed out a decrease in CuAl₂ tensile strength as well as an increase in the ductility of this phase at temperatures greater than 150°C, which is approximately 50 percent of the melting point of this intermetallic phase on the absolute temperature scale. These published data have been compiled in Figure 2. The figure includes the hardness of Al-7.86 percent Cu and Al-1.69 percent Cu alloys⁴¹ and also the
relative hardness \( \frac{H_{CuAl_2}}{H_{Al}} \) of 1.69 percent Cu. This relative hardness is not a strong function of temperature and it exhibits a maximum at about 300°C.
II. EXPERIMENTAL PROCEDURE

(a) Melting and Casting

(1) Al-4.5 percent Cu alloy. A unidirectionally solidified ingot of Al-4.5 percent Cu, dimensions 4"x4"x8", was cast under conditions previously described. The mold was heated to 1350°F before pouring the melt. The bottom chill was stainless steel, water-cooled after the melt was poured. The melt was prepared from high purity virgin aluminum and from Al-50 percent Cu master alloy.

Melting was done in a silicon carbide crucible; all tools were coated with zircon wash to minimize iron contamination; degasing was by bubbling chlorine through the melt.

Chemical analysis of the cast ingot revealed a copper content of 4.40 weight percent.

(2) 7075 alloy. A similar unidirectionally solidified ingot of high purity 7075 alloy was previously prepared in this laboratory. The final measured melt composition (in weight percent) was:

\[
\begin{array}{cccccccc}
\text{Zn} & \text{Mg} & \text{Cu} & \text{Cr} & \text{Ti} & \text{Fe} & \text{Si} & \text{Al} \\
5.70 & 2.28 & 1.35 & 0.18 & 0.15 & <0.002 & <0.012 & \text{balance}
\end{array}
\]

(b) Thermomechanical Treatments

(1) Al-4.5 percent Cu alloy. Four slabs of 1/4" thickness were cut perpendicular to the heat flow direction, at distances varying from 1 1/4" to 2 1/2" from the chill surface. Each slab was encased by welding between two other slabs of similar Al-4.5 percent Cu alloy,
as shown in Figure 3. This had the object of reducing heat losses during hot-rolling and handling of the samples. Temperature was monitored by introducing a thermocouple about 3/8" into one side of the center slab.

These four samples were rolled at 900°F, 700°F, 500°F, and room temperature in a Standard Machinery Company rolling mill, with a roll diameter of 6 inches. Each rolling pass produced a reduction in thickness of about 0.020", and the samples were re-heated to rolling temperature in an air circulating furnace after every pass.

In order to obtain the final reduction in thickness of about 8:1, the slabs needed at least three new encasings. No intermediate anneals were necessary for rolling the samples at 900°F and 700°F, but 3 and 7 anneals at 750°F for 1/2 hour were given to the samples rolled at 500°F and room temperature respectively, in order to reach the final 8:1 reduction without cracking.

Specimens with approximately 2:1, 4:1, and 8:1 reduction in thickness were solutionized at 535°C (995°F) for various lengths of time from 0 to 250 minutes in an air circulating furnace and quenched in water. The air circulating furnace had a "bottom drop" arrangement to achieve minimum delay time between solution treatment and quench. Table I summarizes slab locations and deformation conditions.

(2) 7075 alloy. Three slabs of 3/8" thickness were cut perpendicular to the heat flow direction, two of them at 5" and the other at 7" from the chill surface. These slabs were encased by
welding between two other slabs of commercial 7075 alloy sheet, 1/4" thick, in the same manner as described above.

Two samples taken from 5" and 7" from the chill were rolled at 860°F, using a procedure similar to that employed for rolling Al-4.5 percent Cu, to a final reduction of 4:1 and 5:1, respectively. The third sample was solutionized for 12 hours at 860°F and then rolled at 500°F to a final reduction of 4:1.

All samples needed three new encasings and the sample rolled at 500°F needed also three intermediate anneals at 860°F for 1/2 hour in order to obtain the final reduction. Specimens with 2:1, 4:1 and 5:1 reduction in thickness were sealed in Vycor tubings containing helium atmosphere (to avoid loss of zinc), solutionized at 860°F for various lengths of time from 0 to 48 hours and quenched in water.

Another slab, taken at 2" from the chill surface of the ingot was homogenized at 860°F for 12 hours and cold rolled to 4:1 reduction. This sheet provided samples for comparison with the hot and warm rolled material and with the cold rolled material obtained by Singh and Flemings.7,8

(c) Mechanical Testing

Test bars were taken both longitudinally and transverse to the rolling direction. At least three bars (in most cases four) were tested for each thermomechanical treatment on an Instron machine at a strain rate of 0.05 inches per minute; the elongation was measured using a 1/2" extensometer.
(1) **Al-4.5 percent Cu alloy.** The test bar design is shown in Figure 4. After machining the rolled material, the test bars were solution treated at 995°F, quenched and aged at 310°F for 14 hours.

(2) **7075 alloy.** The test bars were machined from rolled material like those of Al-4.5 percent Cu, but with a thickness of 0.060". The test bars were sealed in Vycor tubings containing helium atmosphere, solution treated at 860°F, quenched and aged at 250°F for 24 hours.

(d) **Structural Measurements**

Measurement of the volume percent of second phase was done by quantitative metallography. A two-dimensional systematic point count was used, following the procedure of Hilliard and Cahn. A systematic array of points was used, provided by the corners of a two-dimensional lattice of an 8 x 8 grid. This grid was put in the reticle of the eye piece of an optical microscope. A coarse-mesh lattice criteria was satisfied, using a magnification of 1000 times. When the amount of second phase was greater than about 0.5 percent, approximately 12,000 points were placed on a polished section of the specimen. The fraction of 12,000 points falling on the second phase gave directly the volume fraction of the phase. When the amount of second phase was about 0.5 percent or less, approximately 18,000 points were placed. Table II is a useful guide to determine the approximate number of points to be used; inside each box is given the percent of relative error ($\sigma/V_p$).
and the 67 percent confidence interval (volume fraction = $\sigma$).\textsuperscript{44}

Grain size measurements were made by running a number of random traverses across photomicrographs of the samples and obtaining the mean lineal intercept (Heyn's procedure\textsuperscript{45}). The lineal intercept method was also used to determine the band spacing distribution in rolled 7075 alloy and to measure the particle size in cast Al-4.5 percent Cu.

A metallographic study was made of polished and etched structures before and after mechanical testing. Two different etchants were employed, the chemical compositions are given in Appendix A.

Scanning electron microscopy was also used extensively to study the polished surfaces and to examine directly the fractured surfaces.
III. SOLUTION KINETICS OF Al-4.5 PERCENT Cu ALLOY

(a) **Cast Material**

The structure of the Al-4.5 percent Cu ingot has been examined previously in detail.\(^{46,47}\) Figure 5 shows the non-equilibrium CuAl\(_2\) second phase in a longitudinal section taken at 2" from the chill. Most particles exhibit a very fine eutectic-like structure. A small number of them are in the form of more or less continuous plates parallel to the heat flow direction in the cast ingot (Figure 5b). Figure 6 shows the structural changes that take place during solution treatment. The measured amounts of second phase remaining in the sample after various times of solutionizing at 535°C are plotted in Figure 7. The rate of dissolution becomes small at late stages of heat treatment; even after four hours as much as 0.18 volume percent remains undissolved.

(b) **Effects of Thermomechanical Treatments on Structure**

The intermetallic compound CuAl\(_2\) is more than ten times harder than the Al-rich matrix in Al-4.5 percent Cu at the temperatures at which the alloy was rolled (Figure 2). Therefore, one would expect little deformation of the hard particles while the soft matrix flows around them. Figure 8 to Figure 11 show this. Most particles remain undeformed, although a small number of them break down into smaller ones. It seems that the regions of continuous plates of CuAl\(_2\) phase in cast material (Figure 5b) are those in which most of the breaking down occurs (Figure 12). In the material rolled
at room temperature and at 500°F, several particles fracture even when the amount of deformation is only 38 percent (Figure 13). However, no particle cracks were visible in the samples rolled at 700°F and 900°F. It was observed that rolling at 900°F to 84 percent reduction, noticeably increases the non-uniformity of the second phase particle distribution (Figure 14), agglomerating them in some regions while leaving large regions in the matrix without second phase particles. These scanning electron photomicrographs (Figures 13 and 14b) also reveal that the second phase particles are not massive, but exhibit a eutectic-like structure, composed of CuAl₂ plus Al-rich matrix.

In order to describe quantitatively the rolled structures, as well as the cast structure, it is of interest to calculate the number of second phase particles per unit volume, \( N_V \), and the effect of rolling on this parameter. Assuming the second phase particles to be spheres of equal size, \( N_V \) is given by:

\[
N_V = \frac{\pi}{2} \frac{N_A^2}{N_L}
\]

where:

\( N_L \) = number of particles intersected by a random line of unit length in a random plane of polish

\( N_A \) = number of particles per unit area in a random plane of polish.

Table III shows that the rolling operation results only in a small increase of \( N_V \) (less than 15 percent), compared to undeformed
cast material.

Another characteristic of the cast and the rolled material is that the structures are oriented. The unidirectionally cast structure exhibits "linear orientation", with plane elements parallel to one line (the direction of solidification). Therefore, measurements of $N_L$ will be dependent upon the orientation of the test line. Table IV lists $N_{L||}$ and $N_{L\perp}$, the number of particles intersected by lines parallel and perpendicular to the direction of solidification, respectively. A measure of the degree of orientation of the structure can be obtained by the ratio $N_{L\perp}/N_{L||}$.50

Similarly, for rolled material $N_L$ can be measured in three mutually perpendicular directions. In Table IV, $x$, $y$, and $z$ refer to directions parallel to the rolling direction, perpendicular to the rolling direction but parallel to the rolling plane, and perpendicular to both the rolling direction and rolling plane, respectively. Again, a measure of the degree of orientation is given by $N_{L_y}/N_{L_x}$ and $N_{L_z}/N_{L_x}$.

(c) Solution Kinetics of Rolled Material

The dissolution of the second phase particles during solution treatment is illustrated in Figure 15 for material rolled at four different temperatures; and Figure 16 to Figure 19 show the same quantitatively.

It is interesting to notice that the 84 percent reduction given to the material rolled at 900°F does not result in faster dissolution of CuAl$_2$, whereas the largest reductions given to material
rolled at room temperature, 500°F and 700°F decrease somewhat the
times required for dissolution. Incidentally, the rates of dis-
solution of CuAl₂ in material with least amount of reduction (33 to
39 percent) are identical to those of as-cast material.

Table III showed that the breaking down of some second
phase particles by rolling increased Nᵥ by a small amount. It is
therefore expected that the times required for dissolution of the
second phase would be reduced also by a small amount. Figures 16
through 18 confirm this.

The number of second phase particles per unit volume
was also obtained for cast and rolled material, after 20 minutes
and 150 minutes of solution treatment. This is shown in Table V.
Nᵥ was found to decrease by a small amount after solutionizing for
20 minutes, but it is substantially smaller after solutionizing
for 150 minutes than the value of Nᵥ for material not solutionized.
This explains why the rate of dissolution of second phase decreases
at long times of solution treatment. Also, during the course of
solution treatment, the decrease in Nᵥ is larger for material rolled
at 900°F than for material rolled at the lower temperatures. This
effect could account for the experimental observation that large
reductions at 900°F do not result in faster dissolution of CuAl₂.

(d) Discussion of Analyses of Solution Kinetics

Several analyses which describe solution kinetics of cast
structures containing a non-equilibrium second phase have recently
been given. Singh and Flemings⁷ considered the case of platelike
dendrite morphology, having an initial sinusoidal composition distribution in the primary phase, constant second phase composition, and constant dendrite arm spacing, $l_0$. Precipitate dissolution was assumed to be limited by diffusion of solid in the matrix and the effect of interface motion on solution kinetics was neglected. The expression derived which relates volume fraction of the second phase to solution time is:

$$\frac{g + a}{g_0 + a} = \exp \left( -\frac{Dl_0^2}{4} \right)$$

where:

- $g$ = volume fraction of second phase at time, $t$
- $g_0$ = volume fraction of second phase at time, $t = 0$
- $D$ = diffusion coefficient at the solution temperature
- $l_0$ = one half secondary dendrite spacing
- $t$ = time of solution treatment
- $a = \frac{C_M - C_o}{C_\beta}$
- $C_M$ = maximum solubility at solution temperature
- $C_o$ = initial composition
- $C_\beta$ = composition of $\beta$ phase (non-equilibrium second phase)

Singh, Bardes and Flemings\textsuperscript{51} developed a numerical analysis technique, which employs a matrix solute distribution similar to initial dendrite microsegregation. A platelike morphology was assumed, and the analysis also allows for the effect of interface motion.

A major criticism of these mentioned analyses is the assumption that continuous plates of divorced second phase are located
at the interdendritic regions. From Figure 5, it is clear that most of the second phase is not continuous.

Singh, Bardes and Flemings also discuss in detail the treatment of Teleshov and Zolotorevsky and the numerical solution of Tanzilli and Heckel for dissolution of second phase. Analyses of precipitate dissolution such as those of Thomas and Whelan and Aaron are not applicable to the solution kinetics of Al-4.5 percent Cu because they do not account for overlapping diffusion fields.

Recently, Nolfi, Shewmon and Foster presented an analysis of dissolution and growth kinetics of spherical precipitates, based on the important work of Ham on the theory of precipitation. They developed analytical expressions for the kinetics of dissolution of spherodized, solute-rich precipitates and considered both diffusion-controlled and interface-controlled processes, as well as mixed control in a single development. The following assumptions were made:

1. Spherical precipitates are of equal radii, \( r_0 \).
2. The distribution of precipitates in the matrix is uniform.
3. The matrix is mathematically divided into identical, cubic cells. The precipitate particles are located at the center of each cell. To simplify the problem, they adopted spherical cells of radius \( r_s \) with the same volume as the cubic cells.
4. The diffusion coefficient, \( D \), of solute in the matrix
is independent of time and composition.

(5) The initial solute concentration in the matrix is constant.

(6) Phase boundaries are stationary.

For diffusion controlled dissolution, the results of Ham (and Nolfi, Shewmon and Fouter) can be written:

\[
\frac{\bar{C}_o(t) - \bar{C}_{\alpha_0}}{C_M - \bar{C}_{\alpha_0}} = \frac{6b^2}{1-b^3} \sum_{n=0}^{\infty} f(a_n) \left[1 - \exp(-t/\tau_n)\right] \tag{1}
\]

where:

\[
f(a_n) = \frac{1 + a_n^2}{a_n^2 \left[(1-b) a_n^2 - b\right]}
\]

\(a_n\) is determined from:

\[a_n (1-b) = \tan^{-1}(a_n) + n\pi\]

\[n = 0, 1, 2, \ldots \infty\]

\[b = \left(g_o\right)^{1/3} = \frac{r_o}{r_s}\]

\[\tau_n = \frac{r_s^2}{a_n D}, \text{ or } t/\tau_n = a_n^2 \frac{D}{r_s^2}\]

\[r_o = \text{precipitate radius at time, } t = 0\]

\[r_s = \text{radius of spherical cell (diffusion distance)}\]

\[\bar{C}_\alpha(t) = \text{mean solute concentration in the matrix at time, } t\]

\[\bar{C}_{\alpha_0} = \text{mean initial solute concentration in the matrix (at time, } t = 0)\]
\( C_M \) = solute concentration in the matrix in equilibrium with second phase particles

\( D \) = diffusion coefficient of solute in matrix

\( g_o \) = volume fraction of second phase at time, \( t = 0 \)

To express the left side of equation (1) as a function of \( g \), the volume fraction of second phase at time \( t \), it is necessary to make a mass balance of solute:

\[
\bar{C}_o(t) (1-g) + C_\beta g = C_o \tag{2}
\]

where:

\( C_\beta \) = solute concentration in second phase particles

\( C_o \) = overall initial concentration of solute

At time \( t = 0 \), equation (2) can be written as:

\[
\bar{C}_o (1-g_o) + C_\beta g_o = C_o \tag{3}
\]

Combining equation (2) and equation (3) into equation (1):

\[
\frac{(C_\beta - C_o) (g_o - g)}{(1-g) [C_M - C_o + g_o (C_\beta - C_M)]} = \frac{6b^2}{1-b^3} \sum_{n=0}^{\infty} f(a_n) [1-\exp(-t/\tau_n)]
\tag{4}
\]

Equation (4) is plotted in Figure 20 in the dimensionless form \( g/g_o \) versus \( \frac{Dt}{r_s^2} \). In order to plot the experimental data obtained in Figure 7 into Figure 20, values of \( D \) and \( r_s \) must be known. A value of \( D \) of \( 1.2 \times 10^{-9} \text{cm}^2/\text{sec} \) was used. \(^{60}\) To measure \( r_s \), the procedure
described by Nolfi, Shewmon and Foster was employed. This method consists of measuring the number of second phase particles intersected by a random line of unit length \(N_L\) and the number of second phase particles per unit area in a random plane of polish \(N_A\). Assuming spherical shape and constant particle size, the second phase radius, \(r_0\), is given by:

\[
r_0 = \frac{1}{\pi} \frac{N_L}{N_A}
\]

And if the second phase distribution is uniform,

\[
g_0 = \frac{4/3 \pi r_0^3}{4/3 \pi r_s^3} = \left( \frac{r_0}{r_s} \right)^3
\]

or,

\[
r_s = r_0 \left( g_0 \right)^{-1/3}
\]

Table VI summarizes the values of \(r_s\) obtained. The important point to notice here is that the rolling operation succeeds in reducing the value of \(r_s\) only by a very small amount, compared to the value of \(r_s\) of cast material. This small decrease of the diffusion distance \(r_s\) with rolling explains, through equation (4), the experimental observation that rolling Al-4.5 percent Cu slightly reduces the times required for dissolution of the second phase, except for material rolled at 900°F.

The experimental data of solution kinetics of cast and rolled material is replotted in Figure 20 in the dimensionless form \(g/g_o\).
versus $\bar{D}t/r_s^2$. The agreement with the analysis of Ham and Nolfi, Shewmon and Foster is not very good. On one hand, the analysis predicts more rapid removal of solute in the early stages of solution treatment; this is to be expected because the initial boundary condition assumes constant solute distribution in the matrix (no microsegregation). On the other hand, the assumption of stationary phase boundaries introduces a new error. On the right side of equation (4), appears $b^2$ which was taken as a constant ($b = r_o/r_s$), but actually decreases during solution treatment. Therefore, the assumption of constant $b$ results, again, in faster dissolution of the second phase than in the real situation. Nolfi, Shewmon and Foster\textsuperscript{57} also recognize that assuming constant precipitate size is the major potential source of error and point out that a solution for varying particle size is obtained by substituting $r_o(t)$ for $r_o$. They conclude that such a procedure is applicable though difficult and requires the use of numerical methods and a digital computer.

The analysis of Nolfi, Shewmon and Foster\textsuperscript{57} assumes constant $r_s$, but this was found not to be true for Al-4.5 percent Cu; some second phase particles solutionize faster than others. The diffusion distance $r_s$ can be measured during the course of solution treatment in the same manner as described above. From equation (5) we can write:

$$r_s(t) = r(t) (g)^{-1/3}$$
where \( r(t) \) is the precipitate radius at time \( t \) and \( r_s(t) \) is the radius of the spherical cell (diffusion distance) at time \( t \).

Table VII lists the values of \( r_s(t) \) after solutionizing for 20 to 150 minutes. Tables VI and VII show that the effective diffusion distance, \( r_s \), increases during the course of solution treatment in the same manner as in the case of platelike dendrite morphology.\(^7\)

This effect accounts for the larger deviations, at large values of \( \frac{Dt}{r_s^2} \), of the experimental data from the described analysis, which assumes constant \( r_s \).

It is very significant to notice that the platelike dendritic model of Singh, Bardes and Flemings,\(^51\) also plotted in Figure 20, lies reasonably close to the spherical precipitate model of Ham and Nolfi, Shewmon and Foster, when using the same value of the diffusion distance (that is, using half of a linear spacing between precipitates, \( l_0 \), in place of \( r_s \)).
IV. SOLUTION KINETICS OF 7075 ALLOY

(a) Effects of Thermomechanical Treatments on Structure

The structural features of the cast ingot have been examined previously in detail. Figure 21 shows the typical microstructure, at 5 inches from the chill. It can be seen that most of the non-equilibrium second phase appears at the equiaxed grain boundaries and some of it is present at the interdendritic regions within the grains.

It was observed that even a small amount of deformation by rolling (2/1 reduction), broke down the second phase into small particles, Figure 22. When the amount of reduction was about 4/1, these small particles were arranged in bands oriented in the direction of rolling (Figures 23 to 25); these bands are well defined in the material rolled at 860°F (Figure 23). The samples rolled at 500°F and at room temperature were solutionized for 12 hours before reduction, therefore contain less amount of second phase. For this reason, the bands of second phase are not too well defined (Figures 24 and 25) and, in some regions, no bands can be distinguished (Figure 26). Figures 27 to 30 show the grain structure of rolled material. Material rolled at room temperature shows a completely recrystallized structure after a 52 hour anneal at 860°F. Material rolled at 500°F is partly recrystallized, and material rolled at 860°F does not recrystallize at all and exhibits a very fine sub-grain structure. It is interesting to notice that the second phase particles are located at the original, deformed grain
boundaries (Figures 27 and 28).

Singh and Flemings\(^7\) cold rolled 7075 alloy to larger reduction. Figure 31 shows that at 16/1 reduction, the bands of second phase are made of smaller particles which exhibit less continuity than those in the material rolled 4/1.

A second phase impurity content of 0.55 volume percent was measured; Figure 32 shows some of these impurity particles.

Figure 33 shows the electron microprobe trace of copper concentration versus distance for material rolled 4/1 at 860°F. Notice that the copper content is very large at the regions where the probe trace intersects the bands of second phase particles.

(b) **Effects of Thermomechanical Treatments on Solution Kinetics**

The dissolution on the non-equilibrium second phase particles during solution treatment is illustrated in Figure 34 for material rolled at 860°F and 500°F; Figure 35 and 36 show the same quantitatively. The times required for dissolution of a given amount of second phase are considerable less than those for cast material. Also, increasing the amount of reduction from 2/1 to 4/1 results in faster dissolution of the second phase. This effect is small for the material rolled at 500°F; notice that the initial amount of second phase is very small, about 0.4 volume percent, because of the solution treatment given to this sample before rolling. For a given amount of reduction, and similar initial amounts of second phase, the solution kinetics of material rolled at 860°F are in very good agreement with the reported values of
material cold rolled by Singh and Flemings. But the material rolled at 500°F cannot be compared with the material of Singh and Flemings because of the smaller initial amount of second phase reported here (this is probably due to different annealing treatments during rolling).

Singh and Flemings adopted a physical model to explain the solution kinetics of cast and cold rolled 7075 alloy. They assumed a simple plate-like dendrite morphology, with all the non-equilibrium second phase located at the interdendritic regions. They also assumed the initial concentration distribution within the dendrite arm to be a portion of a sine curve. Precipitate dissolution was limited by diffusion of solute in the primary phase and the motion of the matrix-second phase boundary was neglected. When heat treatment was at a temperature very close to the solvus, they arrived to the following expression:

\[
\frac{g}{g_0} = \exp \left[-\frac{\pi D^2 t}{4 l_0^2}\right]
\]

(6)

where:

- \(g_0\) = volume fraction of undissolved alloy second phase at time, \(t = 0\)
- \(g\) = volume fraction of undissolved alloy second phase at time, \(t\)
- \(D\) = diffusion coefficient of solute in primary phase, \(cm^2 sec^{-1}\)
- \(t\) = time, seconds
- \(l_0\) = one half the dendrite arm spacing, \(cm\)
To examine the effect of working on solution kinetics, they wrote an expression similar to equation (6) for the wrought structure, assuming uniform plate-like bands of divorced second phase, and a sinusoidal concentration distribution within the bands of primary phase:

\[
\frac{g'}{g_0'} = \exp \left[ - \frac{D^2 t}{4 l_0'^2} \right]
\]  

(7)

where:

\[ g' \] = volume fraction of second phase in worked material at time, \( t \)

\[ g_0' \] = Volume fraction of second phase in worked material at time, \( t = 0 \)

\[ l_0' \] = one half the effective diffusion distance in the worked structure.

This model provided a good qualitative description of the solution kinetics of worked material; however, the validity of the model was not tested (i.e., \( l_0' \) was not measured).

To obtain a measurement of the effectiveness of working in reducing the times of solutionizing, it is possible to apply equation (7) of the uniform plate-like model to the present case of bands non-uniformly spaced. This was done by taking \( l_0' \) as one half the apparent uniform band spacing that results in identical rate of dissolution of alloy second phase as the structure described by the non-uniform distribution of bands containing the second phase. Notice that \( l_0' \) is time dependent; it becomes larger as the time of solutionizing increases. This is necessarily so in
oder to account for the large initial rate of dissolution of second phase bands of small spacings; bands of larger spacings become important after solutionizing for longer times, therefore increasing $l_0'$ with time. Figure 37 shows the data for the worked structures, plotted as $\log g'/g'$ versus time. It can be seen that (1) the initial slope of the curves increases with reduction ($l_0'$ decreases with reduction) and (2) the initial slope increases with increasing the initial amount of second phase, for a given amount of reduction.

Following a procedure described by Singh and Flemings, by dividing the logarithm of equation (7) by that of equation (6) yields:

$$\frac{\log (g'/g')}{\log (g_0/g)} = \frac{l_0}{l_0'}$$

Thus, the ratio of the slopes of the curves of Figure 37 (at time approaching zero) yield directly $(l_0/l_0')^2$, a measure of the effectiveness of working in reducing effective diffusion distance and increasing solution rate. This is shown in Figure 38. In very good agreement with Singh and Flemings, this plot indicates that the increase in solution rate is not only a function of the amount of reduction, but also depends on the heat treatment given prior to reduction.
(c) Analysis of Solution Kinetics for Banded Segregates.

Variable Band Spacing

In the present investigation, equation (7) is modified in order to allow for a distribution of band spacings. Figure 39 is a schematic representation of the changes in composition with distance between bands, for different times of solution treatment at the solvus temperature.

The following assumptions are made:

(1) All non-equilibrium second phase is located in parallel bands of uniform composition.

(2) Precipitate dissolution is limited by diffusion of solute into the matrix.

(3) Solute cannot cross a volume element boundary (as defined in Figure 39).

(4) Motion of the matrix-second phase boundary is neglected.

(5) Initial concentration distribution within the matrix, between bands of second phase, are portions of a sine curve.

(6) Temperature of solution treatment is solvus temperature.

Each individual volume element now behaves exactly like those of the Singh and Flemings model. For an element of length $2 l_n$ we have:

$$\frac{g_n}{g_0} = \exp \left[ -\frac{\pi^2 Dt}{4 l_n^2} \right]$$  \hspace{1cm} (8)
where:

\[ g_n = \text{volume fraction of undissolved alloy second phase remaining in element of length } l_n \text{ after time, } t \]

\[ g_o = \text{volume fraction of undissolved alloy second phase at time, } t = 0 \]

(the same for each volume element).

Considering a frequency distribution of band spacings, the overall solution kinetics is given by:

\[
g/g_o = \frac{n_1}{n_t} \exp \left[ -\frac{\pi^2 Dt}{4 l_1^2} \right] + \frac{n_2}{n_t} \exp \left[ -\frac{\pi^2 Dt}{4 l_2^2} \right] + \ldots + \frac{n_n}{n_t} \exp \left[ -\frac{\pi^2 Dt}{4 l_n^2} \right]
\]

or, \( g/g_o = \sum_{i=1}^{n} \frac{n_i}{n_t} \exp \left[ -\frac{\pi^2 Dt}{4 l_i^2} \right] \) \hspace{1cm} (9)

where \( n_i/n_t \) is the fraction of bands that are of spacing \( l_i + \Delta l \).

Figure 40 shows the frequency distributions, \( n_i/n_t \) plotted versus 2 \( l_i \), for material hot rolled 4/1 at 860°F, and material rolled 4/1 and 16/1 at room temperature. The last two samples were selected from the material studied by Singh and Flemings.⁷

Finally, a value of \( D \) must be adopted to make all the terms on the right side of equation (9) known. There is no data available on the diffusion coefficient of Cu in Al-Zn-Mg matrix. \( D \) was calculated from the data of the solution kinetics of the material rolled 4/1 at 860°F (Figure 35) as follows:

Expanding in series the exponential function in equation (9):

\[
g/g_o = \sum_{i=1}^{n} \frac{n_i}{n_t} \left[ 1 - \frac{\pi^2 Dt}{4 l_i^2} + \left(\frac{\pi^2 Dt}{4 l_i^2}\right)^2 \frac{1}{2!} - \left(\frac{\pi^2 Dt}{4 l_i^2}\right)^3 \frac{1}{3!} + \ldots \right]
\]
Taking the derivative respect to time, at time approaching zero becomes:

\[
\lim_{t \to 0} \frac{d(g/g_0)}{dt} = -\pi^2 \frac{D}{4} \sum_{i=1}^{n_t} \frac{n_i}{n_t} \frac{1}{l_i^2}
\]

Rearranging

\[
D = -\pi^2 \frac{n}{\sum_{i=1}^{n_t} \frac{n_i}{n_t} \frac{1}{l_i^2}}\frac{d(g/g_0)}{dt}
\]

With the aid of Figure 35 and Figure 40, a value of \(D = 1.5 \times 10^{-10} \text{ cm}^2 \text{ sec}^{-1}\) is obtained. This value of \(D\) is in very good agreement with published data for the binary Al-Cu at 860°F.\(^6\)

Equation (9) is plotted in Figure 41 in the dimensionless form \(g/g_0\) versus \(Dt/l^2\), where \(l\) is the mean band spacing and is obtained from the frequency distribution curves such as those shown in Figure 40. From equation (9) it is clear that the variable band spacing model presented here will be represented in Figure 41 by one curve for each frequency distribution measured. Three cases are shown, together with points measured experimentally: Material rolled 4/1 at 860°F, material rolled 4/1 at room temperature, and material rolled 16/1 at room temperature (the last two samples were provided by Singh and Flemings).\(^7\) Material rolled 2/1 does not exhibit a band structure (Figure 22), therefore, the bands spacings cannot be measured. Finally, Figure 41 also shows the analysis of Singh and Flemings\(^7\) for the wrought structure, given by equation (7).
The analysis of Singh and Flemings predicts slower dissolution at early times and faster dissolution at late times than the variable band spacing model presented here. This is expected because in the latter model, bands with small spacings account for most of the diffusion at early times and bands with larger spacings become active at longer times. The material rolled 4/1 shows very good agreement with equation (9), but the material rolled 16/1 solutionizes faster than what equation (9) predicts. This can be explained by observing the microstructure of the material, Figure 31. The bands of second phase are broken down into very small particles. This interparticle separation is of the order of $10^{-4}$ cm to $10^{-3}$ cm as compared to the mean band spacing of $5 \times 10^{-3}$ cm; diffusion of solute will take place not only between bands but also between the particles in each band, resulting in faster dissolution of the particles, particularly at short times.

From the foregoing, it is concluded that the model presented here, being described by equation (9), shows very good agreement with the experimental measurements for the case of material having an intermediate amount of reduction (i.e., 4/1), and containing a relatively large initial amount of second phase (at least 1.0 volume percent). Material with either small or very large amounts of reduction (i.e., 2/1 or 16/1) are not properly described by the proposed variable band structure model.
V. EFFECTS OF THERMOMECHANICAL TREATMENTS ON MECHANICAL PROPERTIES

(a) Al-4.5 percent Cu Alloy

In Figure 42 to Figure 46 the ultimate tensile strength, yield strength and percent elongation are plotted versus the amount of undissolved alloy second phase, for material rolled at room temperature, 500°F, 700°F, and 900°F. Also included are the tensile properties of cast material, taken at 2 inches from the chill surface of the ingot. The amount of reduction given to the material rolled at each of the four different temperatures is given in the lower portion of the figures. The ultimate and yield strengths as well as the percent elongation increase by decreasing the amount of alloy second phase. It is important to notice that the different amounts of reduction by rolling have an insignificant effect on tensile properties, as compared with the important effect of the amount of alloy second phase present. The transverse properties are as good as longitudinal properties for material containing small amounts of second phase (0.2 to 0.3 volume percent). However, at larger amounts of second phase, strength and percent elongation are lower in transverse specimens than in longitudinal specimens; this effect becomes less important when rolling temperature is increased (for example, in Figure 46, transverse properties are almost the same as longitudinal properties for material rolled at 900°F, containing 1.03 volume percent of alloy second phase).

The data of the foregoing figures are replotted in Figures 47 to 49 to show the effect of rolling temperature on
mechanical properties. The exact volume percent of second phase contained in each sample is given in the lower portion of each figure. For a given amount of second phase, the ultimate and yield strengths decrease slightly by increasing the rolling temperature. This is attributed to an increase in grain size with rolling temperature, ranging from 0.12mm for cold rolled material to 0.42mm for material rolled at 900°F. It was found that the size of the recrystallized grains depends on the rolling temperature, but not upon the amount of reduction (in the range studied here). From these figures it is also clear that the difference between longitudinal and transverse properties is more marked when the rolling temperature is decreased and when the amount of second phase is increased. The ductility of material containing large amounts of second phase increases by increasing the rolling temperature. This is probably due to the presence of interparticle cracks only in material rolled at the lower temperatures, (Figure 13).

It is of interest in this work that ultimate and yield strengths also increase with decreasing amounts of second phase. This was explained by Singh and Flemings for 7075 alloy. An identical explanation pertains to the Al-Cu system: When the second phase dissolves, the matrix becomes richer in solute and therefore, stronger.

The reason why transverse properties in the rolled material are somewhat lower than longitudinal properties was attributed to the fact that the second phase particles are not randomly distributed but are strung out in arrays in the rolling direction.
Scanning electron microscopy of the fractured test bars reveals dimples characteristic of ductile fracture. In samples partially solutionized, some dimples are relatively large and contain second phase particles, Figure 50. But after solutionizing for very long times, the dimples are very small and no second phase particles are visible, Figure 51.

(b) 7075 Alloy

Figures 52 and 53 show the ultimate strength, yield strength and percent elongation plotted versus amount of second phase, for material taken at 5 inches from the chill surface of the ingot and rolled to 4/1 reduction at 860°F and 500°F, and for material taken at 7 inches from the chill and rolled to 5/1 reduction at 860°F. The data obtained by Singh and Flemings for cold rolled material is also plotted in these figures. Strength and ductility increase by decreasing the amount of second phase. The properties measured in the transverse direction are smaller than the properties measured in the longitudinal direction, even when the second phase content is small. The explanation for this, similar to the case of Al-4.5 percent Cu, is based on the structural arrangement of second phase particles parallel to the rolling direction. The strength values determined in this investigation, although slightly larger than those reported by Singh and Flemings, are in close agreement with them. However, like in the case of Al-4.5 percent Cu alloy, the ductility of material rolled at high temperature (860°F) and containing large amounts of second phase is substantially higher than
the ductility of cold rolled material with similar amounts of second phase. In particular, it is interesting that the sample taken from 7 inches from the chill surface of the cast ingot (at only one inch from the ingot top, with a dendrite arm spacing of 100 microns) and rolled at 860°F exhibits larger ductility at high second phase contents than the material taken from closer distances from the chill and cold rolled.  

A possible explanation of the observation that, at large second phase contents, cold rolling lowers ductility lies on the experimental evidence that several second phase particles fracture during cold rolling, whereas no evidence of interparticle cracking was found after rolling at 860°F. On the other hand, Di Russo points out that the fine, stable substructure obtained after hot rolling could also account for an increase in ductility, compared to cold rolled material.

The fracture surface of material not completely solutionized, Figure 54 reveals the important role of second phase particles on fracture, a number of them being clearly seen. Material solutionized longer times exhibits dimples characteristic of ductile fracture (Figure 55). Figure 56 shows the fracture of material solutionized 19 hours. The fracture appearance is of the ductile type; however, a region containing some second phase particles was enlarged to show the contrast between the broken particles and the very small dimples in the matrix around them.

Appendices B and C tabulate the mechanical properties obtained after processing Al-4.5 percent Cu and 7075 alloys.
VI. SUMMARY AND CONCLUSIONS

The purpose of this investigation was to describe qualitatively and quantitatively the structure, solution kinetics and mechanical properties of unidirectionally solidified Al-4.5 percent Cu and 7075 ingots after various thermomechanical treatments. The thermomechanical treatments consisted of deforming sections of the cast ingot to various amounts of reduction by rolling at selected temperatures, followed by solution treatments of several lengths of time.

From the present investigation it is concluded that:

(1) Rolling Al-4.5 percent Cu alloy breaks down only relatively few second phase particles, the Al-rich matrix flows around the particles during deformation. Interparticle spacing (a measure of diffusion distance) is affected only slightly by working.

(2) Rolling 7075 alloy breaks down the second phase into small particles; these particles are arranged in bands. Band spacings (measures of diffusion distances) are markedly reduced by working.

(3) Rolling Al-4.5 percent Cu alloy at 900°F does not result in faster dissolution of CuAl₂ compared to as-cast material, whereas the largest reduction given to material rolled at room temperature, 500°F
and 700°F decreases slightly the times required for dissolution.

(4) Rolling 7075 alloy at 860°F, 500°F and room temperature results in faster solution kinetics than as-cast material. By increasing the amounts of reduction, a decrease in the times required for dissolving a given amount of second phase is observed.

(5) A model is presented for solution kinetics of banded segregates, where band spacing is not uniform. A simple analytic expression is obtained. Quantitative agreement between the analysis and experimental results on wrought 7075 alloy is good.

(6) An analysis of Ham's \(^5\) is applied to dissolution of the approximately spherical second phase particles in Al-4.5 percent Cu alloy. Solution kinetics are slower than predicted by this analysis, especially at long times of solution treatment. The discrepancy between the analysis and experiments occurs presumably because (a) the analysis assumes constant solute distribution in the matrix (no microsegregation), and (b) the analysis neglects the effect of interface movement on solution rate.

(7) In Al-4.5 percent Cu and 7075 alloys, strength and ductility increase when the amount of second phase decreases.

(8) In rolled Al-4.5 percent Cu, transverse tensile
properties are identical to longitudinal properties at low contents of second phase (about 0.2 volume percent in material rolled at 500°F and room temperature, and about 0.5 percent in material rolled at 700°F and 900°F). Also, transverse tensile properties become more similar to longitudinal properties, the higher the temperature of rolling.

(9) Tensile strength and ductility of wrought 7075 alloy measured in the longitudinal direction are always larger than strength and ductility measured in the transverse direction.

(10) 7075 alloy can be rolled at 860°F without prior solution treatment. Tensile and yield strengths of material rolled at 860°F and 500°F are slightly better than the properties previously obtained from material rolled at room temperature. Ductility of material rolled at 860°F and containing large amounts of second phase is substantially higher than the ductility of cold rolled material with similar amounts of second phase.

(11) A section of unidirectionally cast 7075 alloy, taken at 7 inches from the chill surface of an 8 inch long casting (having a dendrite arm spacing of 100 microns), was successfully rolled to 5/1 reduction and exhibited mechanical properties comparable with those of material taken from closer distances from the chill.
TABLE I

Location of Al-4.5 percent Cu Samples and Deformation Conditions

<table>
<thead>
<tr>
<th>Rolling Temperature (°F)</th>
<th>900</th>
<th>700</th>
<th>500</th>
<th>R.T. as cast</th>
</tr>
</thead>
</table>

<table>
<thead>
<tr>
<th>Distance from chill sample taken from, (inches)</th>
<th>1 1/2</th>
<th>1 1/4</th>
<th>1 3/4</th>
<th>2 1/4</th>
<th>2</th>
</tr>
</thead>
</table>

<table>
<thead>
<tr>
<th>Percent reduction of Solution kinetics samples</th>
<th>38</th>
<th>39</th>
<th>33</th>
<th>38</th>
</tr>
</thead>
</table>

<table>
<thead>
<tr>
<th></th>
<th>72</th>
<th>71</th>
<th>73</th>
<th>74</th>
</tr>
</thead>
</table>

<table>
<thead>
<tr>
<th></th>
<th>84</th>
<th>88</th>
<th>87</th>
<th>88</th>
</tr>
</thead>
</table>
**TABLE II**

Percent of Relative Error and Confidence Interval in Volume Fraction Measurements

<table>
<thead>
<tr>
<th>V_V VOLUME FRACTION</th>
<th>TOTAL NUMBER OF POINTS APPLIED</th>
<th>6000</th>
<th>12000</th>
<th>18000</th>
</tr>
</thead>
<tbody>
<tr>
<td>3%</td>
<td></td>
<td>7.5%</td>
<td>5.3%</td>
<td>4.3%</td>
</tr>
<tr>
<td></td>
<td></td>
<td>2.78% to 3.22%</td>
<td>2.84% to 3.16%</td>
<td>2.87% to 3.17%</td>
</tr>
<tr>
<td>1%</td>
<td></td>
<td>13%</td>
<td>9.1%</td>
<td>7.5%</td>
</tr>
<tr>
<td></td>
<td></td>
<td>0.87% to 1.13%</td>
<td>0.91% to 1.09%</td>
<td>0.93% to 1.07%</td>
</tr>
<tr>
<td>0.5%</td>
<td></td>
<td>18%</td>
<td>13%</td>
<td>11%</td>
</tr>
<tr>
<td></td>
<td></td>
<td>0.41% to 0.59%</td>
<td>0.44% to 0.56%</td>
<td>0.45% to 0.55%</td>
</tr>
<tr>
<td>0.3%</td>
<td></td>
<td>23.6%</td>
<td>17%</td>
<td>13.6%</td>
</tr>
<tr>
<td></td>
<td></td>
<td>0.23% to 0.37%</td>
<td>0.25% to 0.35%</td>
<td>0.269% to 0.331%</td>
</tr>
<tr>
<td>0.1%</td>
<td></td>
<td>41%</td>
<td>29.5%</td>
<td>23.6%</td>
</tr>
<tr>
<td></td>
<td></td>
<td>0.059% to 0.141%</td>
<td>0.071% to 0.129%</td>
<td>0.077% to 0.123%</td>
</tr>
</tbody>
</table>

Quantities in Boxes Represent:

\[
\sigma/V_V \\
67\% CONFIDENCE INTERVAL
\]

\(V_V\) = volume fraction of second phase

\(\sigma\) = standard deviation

\(\sigma/V_V\) = percent of relative error
<table>
<thead>
<tr>
<th>Material</th>
<th>$N_L$ (cm$^{-1}$)</th>
<th>$N_A$ (cm$^{-2}$)</th>
<th>$N_V$ (cm$^{-3}$)</th>
</tr>
</thead>
<tbody>
<tr>
<td>cast</td>
<td>30</td>
<td>$1.66 \times 10^4$</td>
<td>$1.44 \times 10^7$</td>
</tr>
<tr>
<td>Rolled 88%</td>
<td>28</td>
<td>$1.71 \times 10^4$</td>
<td>$1.64 \times 10^7$</td>
</tr>
<tr>
<td>at R.T.</td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>Rolled 87%</td>
<td>30</td>
<td>$1.75 \times 10^4$</td>
<td>$1.60 \times 10^7$</td>
</tr>
<tr>
<td>at 500°F</td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>Rolled 88%</td>
<td>27</td>
<td>$1.69 \times 10^4$</td>
<td>$1.66 \times 10^7$</td>
</tr>
<tr>
<td>at 700°F</td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>Rolled 84%</td>
<td>26</td>
<td>$1.60 \times 10^4$</td>
<td>$1.55 \times 10^7$</td>
</tr>
<tr>
<td>at 900°F</td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>Material</td>
<td>$N_{L_{</td>
<td></td>
<td>}}$ (cm$^{-1}$)</td>
</tr>
<tr>
<td>---------------</td>
<td>--------------------------</td>
<td>----------------------------</td>
<td>----------------------------</td>
</tr>
<tr>
<td>Cast</td>
<td>27</td>
<td>34</td>
<td>1.3</td>
</tr>
</tbody>
</table>

<table>
<thead>
<tr>
<th>Material</th>
<th>$N_{L_{x}}$ (cm$^{-1}$)</th>
<th>$N_{L_{y}}$ (cm$^{-1}$)</th>
<th>$N_{L_{z}}$ (cm$^{-1}$)</th>
<th>$N_{L_{y}}/N_{L_{x}}$</th>
<th>$N_{L_{z}}/N_{L_{x}}$</th>
</tr>
</thead>
<tbody>
<tr>
<td>Rolled 88% at R. T.</td>
<td>27</td>
<td>27</td>
<td>30</td>
<td>1.0</td>
<td>1.1</td>
</tr>
<tr>
<td>Rolled 87% at 500°F</td>
<td>27</td>
<td>27</td>
<td>32</td>
<td>1.0</td>
<td>1.2</td>
</tr>
<tr>
<td>Rolled 88% at 700°F</td>
<td>23</td>
<td>26</td>
<td>29</td>
<td>1.1</td>
<td>1.3</td>
</tr>
<tr>
<td>Rolled 84% at 900°F</td>
<td>20</td>
<td>28</td>
<td>33</td>
<td>1.4</td>
<td>1.6</td>
</tr>
</tbody>
</table>
TABLE V

Parameters Measured in Al-4.5 percent Cu after Solutionizing for 20 Minutes and 150 Minutes

<table>
<thead>
<tr>
<th>Material</th>
<th>Time of solutionizing (minutes)</th>
<th>$N_L$ (cm$^{-1}$)</th>
<th>$N_A$ (cm$^{-2}$)</th>
<th>$N_V$ (cm$^{-3}$)</th>
</tr>
</thead>
<tbody>
<tr>
<td>Cast</td>
<td>20</td>
<td>27</td>
<td>$1.54 \times 10^4$</td>
<td>$1.36 \times 10^7$</td>
</tr>
<tr>
<td>Rolled 88% at R. T.</td>
<td>20</td>
<td>19</td>
<td>$1.38 \times 10^4$</td>
<td>$1.55 \times 10^7$</td>
</tr>
<tr>
<td>Rolled 87% at 500°F</td>
<td>20</td>
<td>19</td>
<td>$1.22 \times 10^4$</td>
<td>$1.23 \times 10^7$</td>
</tr>
<tr>
<td>Rolled 88% at 700°F</td>
<td>20</td>
<td>17</td>
<td>$1.24 \times 10^4$</td>
<td>$1.42 \times 10^7$</td>
</tr>
<tr>
<td>Rolled 84% at 900°F</td>
<td>20</td>
<td>20</td>
<td>$1.13 \times 10^4$</td>
<td>$1.01 \times 10^7$</td>
</tr>
<tr>
<td>Cast</td>
<td>150</td>
<td>9.5</td>
<td>$0.67 \times 10^4$</td>
<td>$0.73 \times 10^7$</td>
</tr>
<tr>
<td>Rolled 88% at R. T.</td>
<td>150</td>
<td>9.8</td>
<td>$0.65 \times 10^4$</td>
<td>$0.67 \times 10^7$</td>
</tr>
<tr>
<td>Rolled 87% at 500°F</td>
<td>150</td>
<td>6.0</td>
<td>$0.39 \times 10^4$</td>
<td>$0.40 \times 10^7$</td>
</tr>
<tr>
<td>Rolled 88% at 700°F</td>
<td>150</td>
<td>5.0</td>
<td>$0.39 \times 10^4$</td>
<td>$0.48 \times 10^7$</td>
</tr>
<tr>
<td>Rolled 84% at 900°F</td>
<td>150</td>
<td>7.1</td>
<td>$0.42 \times 10^4$</td>
<td>$0.39 \times 10^7$</td>
</tr>
</tbody>
</table>
**TABLE VI**

Radii of Spherical Cells in Cast and Rolled Al-4.5 percent Cu

<table>
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<tr>
<th>Material</th>
<th>$r_s$ (cm)</th>
</tr>
</thead>
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<tr>
<td>cast</td>
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</tr>
<tr>
<td>Rolled 88% at R. T.</td>
<td>$1.87 \times 10^{-3}$</td>
</tr>
<tr>
<td>Rolled 87% at 500°F</td>
<td>$1.97 \times 10^{-3}$</td>
</tr>
<tr>
<td>Rolled 88% at 700°F</td>
<td>$1.92 \times 10^{-3}$</td>
</tr>
<tr>
<td>Rolled 84% at 900°F</td>
<td>$1.99 \times 10^{-3}$</td>
</tr>
<tr>
<td>Material</td>
<td>Time of Solutionizing (minutes)</td>
</tr>
<tr>
<td>------------------------</td>
<td>---------------------------------</td>
</tr>
<tr>
<td>Cast</td>
<td>20</td>
</tr>
<tr>
<td>Rolled 88% at R. T.</td>
<td>20</td>
</tr>
<tr>
<td>Rolled 87% at 500°F</td>
<td>20</td>
</tr>
<tr>
<td>Rolled 88% at 700°F</td>
<td>20</td>
</tr>
<tr>
<td>Rolled 84% at 900°F</td>
<td>20</td>
</tr>
<tr>
<td>Cast</td>
<td>150</td>
</tr>
<tr>
<td>Rolled 88% at R. T.</td>
<td>150</td>
</tr>
<tr>
<td>Rolled 87% at 500°F</td>
<td>150</td>
</tr>
<tr>
<td>Rolled 88% at 700°F</td>
<td>150</td>
</tr>
<tr>
<td>Rolled 84% at 900°F</td>
<td>150</td>
</tr>
</tbody>
</table>
Figure 1. Method of handling small specimens for hot rolling by Maunder and Charles
Figure 2. Microhardness of CuAl₂ and Al-rich matrix
Figure 3. Method of encasing specimens for rolling
Figure 4. Dimensions of the test bars. (Drawing not to scale)
The microstructure of the as-cast Al-6.5 percent Cu ingot, from a section parallel to the heat flow direction, at a distance from the chill. Magnifications: (a) 150x, (b) 400x.
Figure 6. Structural changes during solution treatment of as-cast Al-4.5 percent Cu. Magnification: 150X.
(a) No solution treatment, (b) solutionized 20 minutes, (c) solutionized 150 minutes.
Figure 7. Undissolved alloy second phase versus time of solutionizing, for as-cast Al-4.5 percent Cu taken from 2 inches from the chill.
Figure H. Microstructure of Al-4.5 percent Cu, rolled 88 percent at room temperature. Notice that most of the second phase particles are not deformed. Rolling direction is horizontal. Magnification: 150X. (a) Longitudinal section, (b) transverse section.
Figure 9. Microstructure of Al-4.5 percent Cu, rolled 87 percent at 500°F. Longitudinal section. Rolling direction is horizontal. Magnification: 150X.

Figure 10. Microstructure of Al-4.5 percent Cu, rolled 88 percent at 700°F. Longitudinal section. Rolling direction is horizontal. Magnification: 150X.
Figure 11. Microstructure of Al-4.5 percent Cu, rolled 84 percent at 900°F. Most of the second phase particles are not deformed. Rolling direction is horizontal. Magnification: 150X. (a) Longitudinal section, (b) transverse section.
Figure 12. Microstructure of Al-4.5 percent Cu, showing some second phase particles broken down. Compare with Figure 1(b). Magnification: 150X. (a) Rolled 88 percent at room temperature. (b) Rolled 84 percent at 900°F.
Figure 13. Scanning electron microscope photographs of Al-4.5 percent Cu, rolled 38 percent at room temperature, showing fractured second phase particles. Rolling direction is horizontal. Longitudinal plane section in (a) and (b). Rolling plane section in (c). Magnifications: (a) 250X, (b) 625X, (c) 600X.
Figure 14. Microstructure of Al-4.5 percent Cu rolled 84 percent at 900°F, showing non-uniform array of second phase particles. Longitudinal plane. Rolling direction is horizontal. Magnifications: (a) 400X, (b) 625X.
Figure 15. Structural changes during solution treatment of Al-4.5 percent Cu. Magnification: 50x.
(a), (b), (c): Rolled 88% at R.T.; (d), (e), (f): Rolled 87% at 500°F. (a) and (d): Treated for 48 hours in solution treatment; (b) and (e): Solutionized 20 minutes; (c) and (f): Solutionized 180 minutes.
Figure 1d. Continued, (g), (h), (i): Rolled 88% at 700°F; (j), (k), (l): Rolled and at 800°F.
(g) and (j): No solution treatment; (n) and (k): Solutionized 2 hours.
(i) and (l): Solutionized 150 minutes.
Figure 16. Undissolved alloy second phase versus time of solutionizing for Al-4.5% Cu rolled at room temperature.
Figure 17. Undissolved alloy second phase versus time of solutionizing, for Al-4.5 percent Cu rolled at 500°F.
Figure 18. Undissolved alloy second phase versus time of solutionizing for Al-4.5 percent Cu rolled at 700°F.
Figure 10. Undissolved alloy second phase versus time of solutionizing. AI - 4.5% Cu rolled 900°F.
Figure 20. Dimensionless amount of undissolved second phase, \( \frac{g}{g_0} \), versus dimensionless time, \( \frac{Dt}{r_s^2} \), for Al-4.5 percent Cu.
Figure 21. Microstructure of cast 7075 alloy, at 5 inches from the chill. Magnification: 88X.

Figure 22. Microstructure of 7075 alloy rolled 2/1 at 860°F. Longitudinal plane. Rolling direction is horizontal. Magnification: 150X.
Figure 23. Microstructure of 7075 alloy rolled at 860°F. Rolling direction is horizontal. (a) and (b): Rolled 4/1. Longitudinal plane. (c) and (d): Rolled 5/1. Rolling plane. Magnifications: (a) and (c): 150X, (b) and (d): 400X.
Figure 23. (continued)
Figure 24. Microstructure of 7075 alloy, rolled 4/1 at 500°F, solutionized 1/2 hour at 860°F. Longitudinal plane. Rolling direction is horizontal. Magnification: 100X.
Figure 25. Microstructure of 7075 alloy, rolled 4/1 at room temperature, solutionized 1/2 hour at 860°F. Rolling direction is horizontal. Magnifications: (a) 150X, (b) 400X.
Figure 26. Microstructure of 7075 alloy rolled 4/1 at 500°F, solutionized 1/2 hour at 860°F. Longitudinal plane. Rolling direction is horizontal. Magnification: 150X.
Figure 27. Grain structure of 7075 alloy, rolled 4/1 at 860°F. Longitudinal plane. Rolling direction is horizontal. Magnification: 150X.

Figure 28. Grain structure of 7075 alloy, rolled 4/1 at 860°F, solutionized 1/2 hour at 860°F. Rolling plane. Rolling direction is horizontal. Magnification: 150X.
Figure 29. Grain structure of 7075 alloy, rolled 4/1 at 500°F, solutionized 52 hours at 860°F. Rolling direction is horizontal. (a) Rolling plane. Magnification: 90X. (b) Longitudinal plane. Magnification: 150X. The darker areas are rich in chromium.
Figure 30. Grain structure of 7075 alloy, rolled 4/1 at room temperature, solutionized 52 hours at 860°F. Longitudinal plane. Rolling direction is horizontal. Magnification: 150X.
Figure 11. Microstructure of 7075 alloy rolled 16/1 at room temperature. Rolling direction is horizontal. Magnification: 150X.  
(a) Longitudinal plane. (b) Rolling plane.
Figure 12. Impurity particles in 7075 alloy. In (a) notice that the angular impurity (probably \( \mathrm{FeAl}_2 \)) is next to a non-equilibrium second phase particle. Magnification: 10,000X.
A line micrographic trace showing copper concentration versus distance for a sample taken at 500°F. Photomicrographic above shows the probe trace.
Figure 36. Structural changes during solution treatment of 2075 alloy. Magnification: 150X. (a), (b), (c), (d): Rolled 4/1 at 300°F; (e), (f), (g), (h): Rolled 4/1 at 800°F. (a) and (c): No solution treatment; (b) and (f): Solutionized 4 hours; (c) and (g): Solutionized 12 hours; (d) and (h): Solutionized 24 hours.
Figure 34. (continued)
Figure 35. Undissolved alloy second phase versus solutionizing time for 7075 alloy rolled at 860°F.
Figure 36. Undissolved alloy second phase versus solutionizing time for 7075 alloy rolled at 500°F. Alloy was solutionized 12 hours prior to reduction.
Figure 37. Plot of log (g'/g) versus solution time for 7075 alloy rolled at 860°F and 500°F.
Figure 38. Plot of \( \left( \frac{l_0}{l'_o} \right)^2 \) versus amount of reduction for 7075 alloy rolled at 860°F and 500°F.
Figure 39. Schematic diagram of composition versus distance for three different times of solution treatment at the solvus temperature.
Figure 40. Frequency distribution of band spacings in 7075 alloy:
(a) Rolled 4/1 at 860°F
(b) Rolled 4/1 at room temperature
(c) Rolled 16/1 at room temperature
Figure 41. Dimensionless amount of undissolved second phase, $g/g_0$, versus dimensionless time, $Dr/T^2$. for 7075 alloy. (a) Material rolled 4:1. (b) Material rolled 16:1.
Experimentally Measured:

\( \Delta : 7075 \) rolled 16/1 at R.T.,
(from Sinph and Flemings)

Analysis of Singh and Flemings equation (7)

Calculated curve for 7075 alloy rolled 16/1 at R.T.

Figure 41. (continued)
Figure 42. Tensile properties versus percent of undissolved alloy second phase, for cast Al-4.5 percent Cu taken from 2 inches from the chill.
Figure 43. Tensile properties versus percent of undissolved alloy second phase, for Al-4.5 percent Cu rolled at room temperature. Percentage in lower part of figure indicate reduction by rolling.
Figure 44. Tensile properties versus percent of undissolved alloy second phase, for Al-4.5 percent Cu rolled at 500°F. Percentages in lower part of figure indicate reduction by rolling.
Figure 45. Tensile properties versus percent of undissolved alloy second phase, for Al-4.5 percent Cu rolled at 700°F. Percentages in lower part of figure indicate reduction by rolling.
Figure 46. Tensile properties versus percent of undissolved alloy second phase, for Al-4.5 percent Cu rolled at 900°F. Percentages in lower part of figure indicate reduction by rolling.
Figure 47. Tensile properties versus rolling temperature, for Al-4.5 percent Cu containing approximately 0.9 volume percent second phase. Percentages in lower part of figure indicate the exact amounts of second phase.
Figure 48. Tensile properties versus rolling temperature, for Al-4.5 percent Cu containing approximately 0.5 volume percent second phase. Percentages in lower part of figure indicate the exact amounts of second phase.
Figure 49. Tensile properties versus rolling temperature, for Al-4.5 percent Cu containing approximately 0.23 volume percent second phase. Percentages in lower part of figure indicate the exact amounts of second phase.
Figure 50. Scanning electron fractographs of Al-4.5 percent Ti showing fractured particle inside large dimple. Second phase particle in (b) exhibits eutectic like structure.
(a) Rolled 45 percent at 900°F and solutionized 30 minutes at 545°C. Magnification: 925X.
(b) Rolled 48 percent at room temperature and solutionized 2 minutes at 545°C. Magnification: 925X.
Figure 51. Scanning electron fractograph of Al-4.5 percent Cu rolled 72 percent at 900°F and solutionized 250 minutes at 535°C. Magnification: 1870X.
Figure 52. Longitudinal tensile properties of 7075 alloy versus undissolved alloy second phase.
Figure 53. Transverse tensile properties of 7075 alloy versus undissolved alloy second phase.
Figure 14: Scanning electron fractograph of 7075 alloy rolled out at 500°F and solutionized 1/2 hour at 500°F. Magnification: 1360X.
Figure 54. Scanning electron fractograph of 7075 alloy rolled 4/1 at 500°F and solutionized 1/2 hour at 860°F. Magnification: 1360X.

Figure 55. Scanning electron fractograph of 7075 alloy rolled 4/1 at 500°F and solutionized 52 hours at 860°F. Magnification: 3000X.

INTENTIONAL DUPLICATE EXPOSURE
Figure 56. Scanning electron fractograph of 7075 alloy rolled 4/1 at room temperature and solutionized 16 hours at 860°F. Magnification: (a) 700X, (b) 2800X.
REFERENCES


Etchants used for Al-4.5 percent Cu and 7075 alloys

(1) Solution of 25 percent HNO₃ at 70°C. Immerse sample for 10 seconds, quench in cold water. This etch reveals the second phase particles with little attack in the matrix.

(2) Solution of 5 drops HF, 10 drops HNO₃, 3 ml lactic acid and 80 ml H₂O. Immerse sample 10 to 40 seconds. This etch reveals the microstructure of 7075 alloy.
APPENDIX B

Mechanical Properties of Al-4.5 percent Cu after processing

<table>
<thead>
<tr>
<th>Rolling Temperature</th>
<th>Percent Reduction</th>
<th>Volume Percent of Alloy Phase</th>
<th>LONGITUDINAL PROPERTIES</th>
<th>TRANSVERSE PROPERTIES</th>
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<tr>
<td></td>
<td></td>
<td></td>
<td>Y.S. (psi x 10^{-3})</td>
<td>U.T.S. (psi x 10^{-3})</td>
</tr>
<tr>
<td>Room Temperature</td>
<td></td>
<td></td>
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<td></td>
</tr>
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<td>0.76</td>
<td>29.2</td>
<td>45.0</td>
<td>8.0</td>
</tr>
<tr>
<td>38</td>
<td>0.45</td>
<td>30.5</td>
<td>48.0</td>
<td>10.7</td>
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<tr>
<td>74</td>
<td>0.23</td>
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<td>500°F</td>
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All mechanical properties reported are the average of at least 3 test bars.
APPENDIX C

Mechanical properties of 7075 alloy after processing

<table>
<thead>
<tr>
<th>Sample Location and Deformation</th>
<th>Volume Percent of Alloy</th>
<th>LONGITUDINAL PROPERTIES</th>
<th>TRANSVERSE PROPERTIES</th>
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<tr>
<td></td>
<td>Phase</td>
<td>Y.S. (psi x 10^{-3})</td>
<td>U.T.S. (psi x 10^{-3})</td>
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<td>69.0</td>
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All mechanical properties are the average of 4 test bars.
BIOGRAPHICAL NOTE

The author first saw light on November 29, 1943, in Buenos Aires, Argentina. From 1956 to 1960 he was a cadet at the Liceo Naval Militar "Almirante Guillermo Brown", where he received high school education as well as navy-military training and graduated, with a gold medal, as Guardiamarina de la Reserva Naval Principal de la Infantería de Marina. From 1961 to 1963 he attended the newly found Instituto Tecnológico de Buenos Aires, majoring in Industrial Engineering; but in September 1963 he was transferred to the Department of Metallurgy, M.I.T. He graduated with the degree of Science Bachelor in September, 1965 and Science Master in February, 1967. Then he joined the Solidification of Metals section as a Research Assistant, and served as a Teaching Assistant during the academic year 1968-69. During the scarce moments of leisure while under the doctoral program, the author engaged in several outdoor activities, photography, bridge, and squash became one of his favorite sports. In 1968 he married German-born Erika.