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FOR THE DEGREE OF  
DOCTOR OF PHILOSOPHY  

of  

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British Columbia, 1957  

M.A.Sc., The University of 
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IN ROOM 201, METALLURGY BUILDING  

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"THE EFFECT OF ALLOYING AND COLD ROLLING
ON THE TEXTURE AND MECHANICAL PROPERTIES
OF MAGNESIUM AND MAGNESIUM-LITHIUM ALLOYS"

ABSTRACT

The effect of Li additions to Mg on the texture and mechanical properties in both the hot and cold rolled condition has been examined.

It was found that using the texture goniometer (Schulz technique) only the surface texture is obtained. As a result average textures were prepared for each alloy. Lithium additions to Mg causes a loss in sharpness of the (0001) texture. No indication of a $<\overline{1}1\overline{2}0>$ directional texture was found. The change in texture was explained successfully on the basis of deformation systems active during rolling.

Cold rolling of the alloys caused a loss in sharpness of the (0001) texture for low Li alloys. In the high Li alloys (6 at. % and 12.4 at. %) a pronounced split occurred. A definite $<\overline{1}1\overline{2}0>$ directional texture was observed on the surface of the cold-rolled low Li alloys but this disappeared in the "average" pole figure. Again the change in texture was explained on the basis of deformation systems active during rolling.

Tensile tests of hot-rolled Mg-Li alloys agreed completely with those of Yoshinaga & Horiuchi (9) but showed some variance with those of Hauser, Landon, and Dorn (8).

Tensile tests of cold-rolled Mg-Li alloys showed appreciable strain hardening and a loss of ductility due to the cold work. The higher the Li content the higher the rate of
strain hardening observed for large increments of strain. The ratio of the transverse to longitudinal tensile properties decreased with increasing Li content. A qualitative explanation of the above was made on the basis of active deformation systems.

Limited success was obtained in attempts to correlate mechanical properties and texture in low Li alloys. No attempt was made for high Li alloys.

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THE EFFECT OF ALLOYING AND COLD ROLLING ON THE TEXTURE AND MECHANICAL PROPERTIES OF MAGNESIUM AND MAGNESIUM-LITHIUM ALLOYS

by

GEORGE CLAUDE WOOTTON

B.A.Sc., University of British Columbia, 1957
M.A.Sc., University of British Columbia, 1959

A THESIS SUBMITTED IN PARTIAL FULFILMENT OF THE REQUIREMENTS FOR THE DEGREE OF DOCTOR OF PHILOSOPHY

in the Department of METALLURGY

We accept this thesis as conforming to the required standard

THE UNIVERSITY OF BRITISH COLUMBIA

September, 1967
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Dow Magnesium Co. Ltd. kindly donated the Mg-Li alloys used and performed a number of chemical analysis on the test alloys.

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I. INTRODUCTION

The marked change in the deformation properties of Mg when it is alloyed with lithium (Li) has led to an interest in the deformation mechanisms active in this alloy system for varying temperatures and Li contents. To date all work on the system has been on alloy single crystals or on polycrystalline material in the hot-rolled condition.

(A) Deformation of Pure Mg

The deformation of pure magnesium has been investigated by a number of workers (1, 2, 3, 4, 5, 6). Hauser, Starr, Tietz and Dorn (1) studied polycrystalline Mg with the thought that the need for five independent deformation mechanisms, required for general deformation conditions, might cause pyramidal slip \(\{10\overline{1}1\} <11\overline{2}0\) to occur in addition to the normal basal slip. Pyramidal slip, which had been found to occur in single crystals of Mg at elevated temperatures, was thought to be the most likely additional deformation mode. They found that even in those grains most unfavourable oriented, basal slip \((0001) <11\overline{2}0\) was the only slip system observed. Twinning which occurred exclusively on the \(\{10\overline{1}2\}\) planes, and grain boundary shearing were the other deformation mechanisms observed at room temperature.

In the discussion of reference (1), Roberts reported that the results of work at Dow Chemical Co. substantiated the work of Hauser et al. However he also observed limited non-basal slip of the form \(\{10\overline{1}0\} <11\overline{2}0\) (type I prism plane). The authors replied that since the publication of their paper they had also observed \(\{10\overline{1}0\} <11\overline{2}0\) slip.

In a later publication Hauser, Landon and Dorn (2) examined polycrystalline Mg specimens deformed at temperatures between 78°K and
They observed that basal slip, (0001) \langle 11\bar{2}0 \rangle, was the major deformation mechanism and that grain boundary shearing, which is quite prevalent at 298°K, is rare at 78°K. However as the incidence of grain boundary shearing decreases the amount of prismatic slip \{10\bar{1}0\} \langle 1\bar{1}20 \rangle increases.

Sheely and Nash (3) have demonstrated that the strain-rate controlling mechanism for basal slip in Mg single crystals at low temperatures is the thermally activated intersection of glide dislocations with dislocations threading the basal plane. They also presented data on the variation in critical shear stress with temperature for basal slip.

Flynn, Mote and Dorn (4) reported that in tests on specially oriented single crystals of Mg, prismatic slip was preceded by twinning and fracturing below 450°K but above this temperature extensive prismatic slip preceded fracture. They have reported the critical resolved shear stress (CRSS) values for prismatic slip for the temperature range of 78° to 800°K. It was noted that although neither the Peierls' mechanism nor the dislocation intersection mechanism can account for the observations, Friedel's theory for cross slip to the prismatic plane was in good agreement.

In Friedel's theory it is assumed that dislocations containing edge components slip with ease on the prism plane until they traverse the entire plane, combine with dislocations of opposite sign or become blocked. Such blocking can occur in the vicinity of screw dislocations which readily dissociate with a decrease in energy into their partials on the basal plane. Continued slip therefore requires continued
recombination of the partials on the basal plane to form screw dislocations on the prismatic plane.

Table 1 is taken from Flynn, Mote & Dorn's (4) paper and summarizes the deformation systems possible in pure Mg.

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<td>Prismatic slip</td>
<td>{1010} (1210)</td>
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Yoshinaga and Horiuchi (6) have examined the characteristics of nonbasal slip in Mg single crystals at various strain rates and test temperatures. Of particular interest is the fact that although basal slip has a wide easy glide region, the nonbasal deformation (principally prismatic slip) was found to work harden rapidly at test temperatures of less than 100°C. Above 200°C the work hardening rate decreases to almost zero and the ductility increases.
Hauser, Landon and Dorn (5) reported that polycrystalline Mg obeys two distinct fracture laws. Over the high temperature range the fracture stress decreases with increasing temperature in a manner similar to the flow stress-temperature relationship, while at low temperatures the fracture stress was independent of the test temperature although highly dependent on the grain size. Some plastic strain (1 to 7%) occurred before brittle fracture in all cases. The stress for brittle fracture increases linearly with the reciprocal of the square root of the mean grain diameter, while over the entire range of temperatures investigated the flow strength was observed to increase linearly with the reciprocal of the square root of the mean grain diameter.

(B) Deformation of Mg-Li Alloys

Having established the general deformation modes and fracture behaviour for pure Mg it is of interest to see what the effect will be of Li additions.

Hibbard, Kearney, Hawley and Burke (7) cold rolled Mg and a Mg + 16.4 atomic percent (at. %) Li alloy until cracking occurred to determine the effect of Li additions on the ductility of pure Mg. They found that the Li alloy had a higher ductility and a higher rate of strain hardening than pure Mg.

Hauser, Landon and Dorn (8) did the first thorough examination of Mg-Li alloys. They tested a range of hot rolled alloys with Li contents of from 2.4 at. % to 14.5 at. % at temperatures of 40°K, 78°K, 195°K and 295°K. They observed:

1. that the low temperature ductility of Mg increases with increasing Li content.
2. that low additions of Li increase the rate of strain hardening of the alloy over that of pure Mg but that larger additions decrease the rate to less than that of pure Mg.

3. that although the fracture strength of pure Mg is independent of temperature and failure is of the brittle type the high Li alloys fail in a ductile manner at all temperatures, exhibit localized necking, and have a fracture strength that is dependent on the test temperature.

The reason given for the above observations is the introduction of prismatic slip \( \{10\overline{1}0\} \langle1\overline{1}20\rangle \), in addition to the normal basal slip \( (0001) \langle1\overline{1}20\rangle \), which results from the decreased \( \sqrt{\alpha} \) ratio in the hexagonal lattice caused by the Li additions to the Mg.

Yoshinaga and Horiuchi (9) extended the work on Mg-Li alloys by examining both polycrystals and single crystals. Their results with polycrystals tested at room temperature agreed with Hauser et al (8) in that the low Li alloy specimens had higher rates of strain hardening and higher tensile strengths than pure Mg, while the high Li alloys had lower rates of strain hardening. They extended the polycrystalline results by testing at temperatures up to 400°C. At 300°C and above the flow stress of the alloys is always higher than pure Mg.

The single crystal results of Yoshinaga and Horiuchi show that although the CRSS for basal slip in pure Mg is almost independent of the test temperature above room temperature, the CRSS for Mg-Li alloys begins to decrease with increase in temperature above about 200°C.
This effect is exaggerated by increasing Li contents. However the amount of solid solution hardening increased linearly with increasing Li content for basal slip. Their results on nonbasal slip (both pyramidal $\{10\overline{1}1\}$ $\langle\overline{1}20\rangle$ and prismatic $\{10\overline{1}0\}$ $\langle\overline{1}20\rangle$) suggested that below $200^\circ$C the CRSS for nonbasal slip was lower for 15.5 at. % Li than for pure Mg while above $200^\circ$C the reverse was true. They found that the reversing temperature for flow stress in Mg and Mg-Li alloy polycrystals also occurs near $200^\circ$C.

Another significant point in these results is that the 0.4 at.% Li alloy shows no increase in elongation with the increase in fracture strength. This is thought to imply that nonbasal slip may be almost inactive at room temperature in this dilute alloy as it is in pure Mg. Therefore the increase in flow stress of the dilute alloy polycrystal observed at room temperature may be due to the basal slip hardening resulting from the addition of Li. In the higher Li alloys the dislocation pile-ups on basal planes may be relieved by nonbasal slip. This also will increase the ductility.

Yoshinaga and Horiuchi conclude by showing that of the various models available for explaining the increase in prismatic slip resulting from Li addition that due to Friedel gives the best qualitative agreement. However, quantitatively it does not completely satisfy the experimental results observed.

Ahmadieh, Mitchell and Dorn (10) tested Mg-Li alloy single crystals specially oriented for prismatic slip at temperatures from $4.2^\circ$ to $650^\circ$K. Their results showed that increasing the Li content of the alloy substantially lowers the CRSS for prismatic slip at low temperatures.
whereas at higher temperatures, increased Li increases the CRSS for slip. In the high Li alloys (12.9% and up) the rate controlling mechanism for the temperature range 4.2° to 300°K was purported to be the Peierls mechanism while from 300°K to 400°K the mechanism may be that due to short range ordering suggested by Fisher. For the 7.9 at. % Li alloy and pure Mg above 450°K the rate controlling mechanism is thought to be the result of cross-slip of dislocation segments lying in screw orientation on the basal planes as proposed by Friedel.

(C) Development of Preferred Orientation During Deformation

During mechanical working metals develop a "texture" or preferred orientation of the grains as a result of the deformation process. The most common way of showing the preferred orientation present in a material is by plotting, on a polar diagram, the relative number of the planes under consideration as a function of their angular relationship to a reference plane, usually the surface of the material. Contour lines are then drawn through regions of equal pole population on the polar diagram to produce a pole-figure. (For a detailed explanation see Appendix I.)

In spite of the interest in the texture of hot and cold worked metals and alloys, relatively few attempts have been made to explain how the deformation mechanisms that are active during the working operation can produce the textures which result.

Although Boas and Schmid (11) and Pickus and Mathewson (12) considered the problem of texture development on the basis of the stable end position for crystals, they do not consider how the crystals reach the stable orientation. Taylor (13) considered that perfectly homogeneous
deformation occurs in the aggregate during working. His treatment is mathematically rigorous and although it can be solved for face-centred cubic materials its application to other crystal structures is impractical.

Calnan and Clews (14) made the first attempt to explain, for the three common crystal structures, how the applied stresses during fabrication can produce a particular end texture. They considered the texture that is produced when a single applied stress causes lattice rotation of the crystals in the specimen. For rolling they assume that the textures produced by the tensile and compressive forces can be considered independently in the development of the final texture. The main concept of the Calnan-Clews method is that the directions of the stress axis acting in any grain may be different from the applied stress axis because of the constraints imposed by the surrounding grains. They call the stress axis in each grain the "effective" stress axis. Considering Taylor's hypothesis of five independent deformation mechanisms being required to maintain continuity from grain to grain Calnan and Clews consider that the effective stress must produce a high enough shear stress on a number of different slip systems to make them all operative at once. They consider that when the pole of the effective stress axis is in the centre of the unit triangle single slip occurs, while if the pole of the effective stress axis is on one of the boundaries duplex slip occurs. In both of these cases grain rotation results. However, when the pole of the effective stress axis is at a corner of the unit triangle multiple slip occurs with no corresponding grain rotation. This method covers FCC, BCC and HCP metals.

Liu (15) follows the general approach of Calnan and Clews
regarding the approximation of the rolling process by two stress axes.

He then extends their treatment by considering, as a criterion for the selection of the active slip systems, that the resolved shear stress on the slip system be high and that dislocations active on them must interact attractively so as to lower the total energy content. Liu only considers FCC metals and alloys.

Dillamore and Roberts (16) have modified the basic assumption made by Calnan and Clews regarding the stress axes. Whereas Calnan and Clews assume the two stresses act independently of one another in developing the texture Dillamore and Roberts consider the stress system to be a biaxial one. Also they contend that although five deformation systems may have to be operative near grain boundaries to preserve continuity from grain to grain only one or two planes need to slip in the centre of the grain. Since the centre of the grain contributes much more to the final texture observed than the grain boundary regions the boundary region can be ignored and the deformation of individual grains in the polycrystal can be considered as being similar to the deformation of single crystals. The importance of cross slip is emphasized and its relationship to the stacking fault energy of the material is noted. Both FCC and BCC metals have been considered using this approach.

The approximation of the stress system operative during rolling by a biaxial stress consisting of a compressive stress normal to the sheet and a tensile stress in the rolling direction, or by independent tensile and compressive stresses is justified only when there is no frictional drag between the rolls and the metal. This is generally the case in cold rolling but is a completely erroneous assumption for the
case of hot rolling. Crane and Alexander (17) have shown that when "sticking friction" occurs, as it may in the hot rolling of Mg, there is a marked difference in the deformation at the surface of the material and in the centre. Therefore any texture predictions for hot rolled material must make allowances for this factor.

(D) Correlation Between Preferred Orientation and Mechanical Properties

Once the pole figure of a textured metal is available a general idea of the relationship of transverse to longitudinal properties may be had by examining the symmetry of the contour lines on the pole figure. However, to date no completely successful technique has been devised whereby the tensile properties of a material can be predicted directly from its pole figure.

Avery, Hosford and Backofen (18) have attempted to correlate the transverse and longitudinal strains in tension to both the slip mechanisms operative and the crystallographic texture prevailing in the test specimens. They examined the variation in the ratio of width-to-thickness strains, termed R values, resulting during tension testing of Mg alloys. Magnesium alloys were used in this study because of the strong textures they develop during hot and cold working. The variation of the R values for a number of alloys was shown to be directly related to the degree to which the texture departed from the "ideal", where (0001) planes are parallel to the rolling plane. This in turn influenced the relative amounts of basal and first-order prism slip. A more detailed consideration of this approach will be given in the discussion section of this thesis.

McDonald and Bakarian (19) have attempted to show that
there is, for heavily textured Mg alloys, a correlation between the ratio of both the transverse to longitudinal yield and the transverse to longitudinal ultimate strengths and the ratio of basal planes oriented in the cross-rolling direction quadrants of the pole figure to the planes in the quadrants 90 degrees away (e.g., rolling direction). They obtained reasonable agreement between the ratios measured.

D. V. Wilson (20) has prepared a review of the present understanding of the relationship between preferred orientation, plastic anisotropy, and practical performance of the material in such applications as deep drawing and stretch-forming.

Dillamore (21) reported that tests to establish the compatibility of the Hill Theory of plastic anisotropy (22) and the crystallographic requirements of deformation established that the theory of anisotropic plasticity should not be used to predict the behavior of material under biaxial stressing based on the strain ratio, $R$, derived from uniaxial tensile tests.

Couling (23) has presented a method for correlating the texture of a material and the ratio of its transverse to longitudinal tensile properties. The method consists of summing the effective resolved shear stress for both longitudinal and transverse directions for a number of diameters of the pole figure for the material. The ratio of the sums is shown in the one case examined to agree very well with the ratio of the yield strengths. The mechanisms of this approach will be examined in more detail in the discussion section of this thesis.

Pole figures have been prepared for a large number of Mg alloys prepared by a variety of methods of fabrication. Roberts (24)
gives a review of the preferred orientation resulting from mechanical working of Mg and its alloys. Although most of the commercial Mg alloys have been examined in both the hot and cold-rolled condition the Mg-Li series has received little notice in the hot-rolled condition and almost none in the cold-rolled condition.

The present work was designed to:

- resolve the differences in the results of Yoshinaga and Horiuchi (9) and those of Hauser, Landon, and Dorn (8) on the variation in tensile strength and ductility with Li content (see Figures 29 & 30).
- establish the effect of cold work on the tensile strength and ductility for several Mg-Li alloys.
- establish the change in the hot-rolled texture of Mg with increasing Li content.
- establish the change in the hot-rolled texture due to cold rolling of several Mg-Li alloys.

Only in the first of these are the results of other workers available for comparison. No other results have been found for the tensile properties of cold-worked Mg-Li alloys and only scanty information on the texture of hot and cold-rolled Mg-Li alloys.
II. EXPERIMENTAL

(A) Alloy Preparation

In the initial test program magnesium, (Mg), and magnesium-lithium (Mg-Li) alloys were prepared by melting and casting high purity Mg in graphite containers and Mg-Li alloys in steel containers. Once the alloys of interest had been established Dow Magnesium kindly prepared the alloys listed in Table 1 by melting and permanent mould casting. Dow also supplied the complete chemical analysis for the alloys (Table 2).

(B) Specimen Preparation

1. Homogenization -

   The homogenization of the alloy was accomplished by allowing long reheat times (15 minutes) after each pass in the initial breakdown stages of hot rolling. That this was successful was established by analysis of sections of alloy removed from the final sheet.

2. Hot Rolling -

   The as-cast billets were hot reduced from 2" thickness to 0.100" thickness (95% reduction in thickness) by the rolling sequence described in Table 2. The rolling was done on a two-high variable speed mill with 4" diameter by 6" wide rolls. The rolling temperatures and reductions per pass are those recommended by Dow for initial breakdown of as-cast material. Because of the limitations of the furnaces available the rolling billets had to be cut into shorter lengths at various stages in the rolling. To guarantee soundness and uniformity of condition in the alloys at the final hot-rolled thickness all the pieces followed the sequence outlined in Table 3.

   No provision is made for heating the rolls during hot rolling.
Therefore, because of the high heat loss with thin sheet, the sheet was reheated after each pass. After the last pass in the hot rolling sequence the sheet was reheated for 5 minutes to remove any "cold-rolling" effects.

3. **Cold Rolling**

The best results were obtained when the cold rolling was held to approximately 2% per pass. Larger reductions caused cracking of the sheet after small total reductions. Although all the figures for reductions per pass and total reductions refer to reduction in thickness they are equivalent to reductions in area because of the small transverse spread during the rolling.

In both the hot and cold rolling the sheet was rolled with the same end forward but was rotated 180° around the rolling direction after each pass.

Sheets of alloy 0.030" thick x 2 3/4" wide x 12" to 18" long were prepared in the hot and cold-rolled conditions by the sequences described in Table 2. Material was cold rolled 15, 30, 40, 50, 60, and 70% for the initial tests. These results indicated that specimens in the hot rolled and 15, 30, and 60% cold-rolled conditions would be sufficient to establish the effect of cold work on the mechanical properties and texture. Little heating was observed during cold rolling because of the small reductions per pass and the control pauses used between passes.

The rolled sheets were marked out in pencil to show the orientation of both the transverse and longitudinal tensile specimens and the texture pieces relative to the rolling direction (Figure 2).
Fig. 1. Specimen orientation on rolled sheet.
Fig. 2. Tensile specimen used. Full size.
The individual specimens were identified as to alloy and condition with a vibrating pencil scriber and the sheet was then sheared into pieces on a mechanical shear.

4. **Tensile Specimen Preparation**

Tensile specimens were initially prepared by both stamping and spark machining. Initial consideration of the two techniques showed that although stamping was the easier method for preparing the tensile specimens there was some possibility of the stamping introducing cold work into the specimen. Therefore both metallographic and mechanical tests were run to establish whether or not this occurred. As a result of these tests all tensile specimens were prepared by spark machining. A detailed examination of the test results will be given in a later section. Specimen dimensions are given in Figure 2.

5. **Pickling**

Because of the existence of a thin surface layer of a more highly developed texture than that of the bulk of the material in rolled sheet (26) and in order to remove the deleterious effects of any surface marks, all tensile specimens had 0.001" to 0.002" removed from their as-rolled surface (i.e. 0.002" to 0.004" on the thickness) before testing. The surface material was removed bypickling in 15% HNO₃ in water. The pickling is uniform and the rate of pickling is reasonably controllable. The texture pieces were examined as-rolled and after pickling.

6. **Annealing**

Annealing was carried out in air for the times noted. Since all times were relatively long, .30 minutes or more, and the specimens relatively thin, it was not necessary to allow for heating times. To
ensure that variations in annealing temperature did not introduce additional variables all specimens that were to be annealed at a particular temperature were annealed together and the temperature checked with a thermocouple. Annealing was carried out after the tensile specimens were spark machined but before pickling.

(C) **Tensile Testing**

All tensile tests were run on an Instron Model test machine at a cross-head speed of 0.01 ipm. Specimen mounting in the test grips was carried out using the load cycling control and the load was kept to less than 2 pounds (≈400 psi) in tension, 0 pounds in compression. Loads were measured to the nearest 0.1 pound and elongations were taken from the chart. This allowed a determination of the elongation when localized necking began, a measurement that is not possible with direct micrometer measurements during the test because of the small size of the tensile specimen. An evaluation of the accuracy of cross-sectional area determinations indicated that tensile results were accurate to ± 100 psi.

(D) **Orientation Testing (Pole Figure Determination)**

The specimens for orientation measurements were cut from the rolled sheet by shearing as described in Section B.3. Orientation measurements were made on a Norelco PW1078 texture goniometer. A detailed description of the technique followed in making the measurements and producing a pole figure is given as Appendix I. The pole figures produced show the X-ray intensity contours relative to 100%. The 100% value is the highest intensity measured in the particular test being run. No attempt was made to compare all the specimens to a random standard because of the number of alloys tested and the difficulty of using one
standard (such as pure Mg) for all alloy levels used. To allow for a
general comparison of the intensity levels of the various pole figures
the maximum intensity based on the x 1 scale is written in the upper right
corner of the pole figure. (see Figure 3). The other relevant data that
are needed for a full understanding of the pole figure is also included
on the figure. The 0.002"/0.004" below the as-rolled surface records
that 0.002" has been removed from the material surface by pickling and
that the X-rays used (CoKα) are effective in diffracting from a depth
of 0.002" below the surface. Details of the calculation used to establish
the depth of penetration is given in the Results section.

Since this was the first series of tests to be run on the
texture goniometer a test run was made on an alloy supplied by Dow
Magnesium Co. Ltd. The pole figure plotted from the results agreed very
well with the one supplied with the alloy by Dow.
III. RESULTS AND DISCUSSION

This section of the thesis will consist of three parts:

(A) Presentation and discussion of texture results
(B) Presentation and discussion of tensile results
(C) Discussion of the effect of texture on the corresponding tensile results.

(A)1. Texture of Hot and Cold-Rolled Mg-Li Alloys

In considering the texture of Mg and Mg-Li alloys a knowledge of the depth of metal being examined in each test is essential.

(1) Depth of X-Ray Penetration

To appreciate the significance of a pole figure for a surface of a metal sheet the depth of metal that is represented by the pole figure must be known. Using the equation for X-ray diffractometry given in Cullity (25),

\[ G_X = (1 - e^{-\mu X \sin \theta}) \]

we can calculate the thickness of metal, \( X \), that produces any fraction, \( G_X \), of the total diffracted intensity. For this we must have values for \( \mu = \) linear absorption coefficient of the metal and \( \theta = \) Bragg angle.

For the present work \( \mu = 104.4 \text{ cm}^{-1} \) and \( \theta = 20.1^\circ \)

Solving the equation for the case where \( G_X = 0.95 \)

\[ 0.95 = (1 - e^{-2(104.4)X \sin 20.1^\circ}} \]

\[ e^{-610X} = 0.05 \]

\[ X = \sqrt[6]{610} = 0.0049 \text{ cm} \]

\[ X = 0.0019 \text{ in.} = 0.002 \text{ in.} \]
In order to appreciate how heavily the pole figure must be weighted towards the surface of the metal we can repeat the above calculation for a $G_v$ value of 50% (0.50) and we obtain

$$X = 0.001 \text{ cm} = 0.0004 \text{ in.}$$

This indicates that although the pole figure represents 0.002 in. of metal 50% of the diffraction is occurring in the top 0.0004 in. and indicates that the pole figure is greatly affected by the extreme surface of the sheet.

Dietrich (26) has shown that in diffraction work using Cu radiation the X-ray beam penetrates 0.003" of Mg before its intensity is reduced to an ineffective level. For Fe radiation the figure is 0.002". Since Co radiation has a penetrating power midway between Cu and Fe it would be reduced to an ineffective level by 0.0025" of Mg. This figure agrees well with our calculated value and with the observation that a marked change results in the pole figures shown in this report by the removal of only 0.002" of material.

Having established that no single pole figure accurately represents the full cross-section of a specimen we must now determine how the texture varies from the surface to the centre of a rolled sheet.

(2) **Texture Variation with Thickness**

i) **Hot Rolled Alloys**

To establish the variation of texture with thickness, tests were run on a 0.5 at. % Li alloy and a 12.4 at. % Li alloy, since these represented the two extremes in Li content. Both basal plane (0002) and pyramidal plane $\{10\overline{1}1\}$ pole figures were prepared. The basal pole figures were of interest since the basal planes are the major deformation planes
Fig. 3. Variation in texture with depth for hot-rolled 0.5 at.% Li for (0001).
Fig. 4. Variation in texture with depth for 12.4 at.% Li for (0001).
Fig. 5. Variation in contour location with depth for 0.5 at.% Li.
Fig. 6. Variation in contour location with depth for 12.4 at.% Li.
while the pyramidal pole figures were used to detect the presence of any preferred \(\langle 11\overline{2}0\rangle\) directional texture.

Figure 3 compares the pole figures for four different levels of the 0.5 at. % Li alloy while Figure 4 compares the pole figures for five levels of the 12.4 at. % Li alloy. It is immediately apparent that there is a pronounced loss in sharpness of texture from the surface to the centre of the rolled sheet. In the case of the 12.4 at. % alloy there is a complete splitting of the 100% peak.

To allow a direct comparison of the change in texture with depth the change in the location of each contour on the pole figure was plotted against the depth below the surface represented by the pole figure (Figures 5 and 6). These graphs show the change in each direction separately since the change in the rolling direction is not always accompanied by a corresponding change in the transverse direction.

A discussion of the reasons for this variation follows Section B.

The pyramidal-plane pole figures are shown in Figures 7 and 8 for the two alloys being examined. Again because of the variation in texture with depth it was necessary to prepare pole figures for several depths.

None of the seven pole figures of Figures 7 and 8 show the six-peaked figure characteristic of a material having a \(\langle 11\overline{2}0\rangle\) texture (see Figure 13 surface for an example of this). However, there is some variation in the general pyramidal plane orientation from surface to centre since both alloys show a 100% region in the rolling direction at the surface but show the 100% region in the transverse direction at the centre.
Fig. 7. Pyramidal pole figures for various depths for 0.5 at.% Li.
Fig. 8. Pyramidal pole figures for various depths for 12.4 at.% Li.
Fig. 9. Variation in texture with depth for 30% cold rolled 0.5 at.% Li for (0001).
Fig. 10. Variation in texture with depth for 60% cold rolled 12.4 at.% Li for (0001).
Fig. 11. Variation in contour location with depth for 30% cold rolled 0.5 at.% Li.
Fig. 12. Variation in contour location with depth for 60% cold rolled 12.4 at.% Li.
Appendix II explains the relationship that exists between the basal and pyramidal pole-figures for the same specimen.

ii) Cold Rolled Alloys

Cold working a material may introduce new deformation modes which in turn will change the texture from that produced by hot working. Therefore, the change in texture with depth was determined for 0.5 at. % Li (Figure 9) and 12.4 at. % Li (Figure 10) alloys. The variation of texture with depth is shown graphically in Figures 11 and 12.

The spread in texture shown by the low alloy material is similar to that shown in the hot-rolled condition. (Compare Figure 3 to Figure 9.)

With the cold rolled 12.4 at. % Li alloy the variation with depth is not as immediately apparent. However, an examination of Figure 12 shows that there is a general spreading of the 10, 25 and 50% contours in the rolling directions and a small contraction of the 25% contour in the transverse direction. The most significant change in this alloy is the change in the location of the 100% peaks at the various depths.

The pyramidal pole figures for the two alloys are shown in Figures 13 and 14. The 0.5 at. % Li alloy shows a definite \(\langle 1120 \rangle\) directional texture on the as-rolled surface. This disappears with the removal of the surface metal. The pole figure for the surface of the 12.4 at. % Li alloy shows a slight \(\langle 1120 \rangle\) texture while the remaining pole figures show a completely random directional texture in spite of their numerous intensity peaks. An explanation of this is given in the Appendix II.
Fig. 13. Pyramidal pole figures for various depths of 30% cold rolled 0.5 at.% Li alloy.
Fig. 14. Pyramidal pole figure for various depths of 60% cold rolled 12.4 at.% Li alloy.
Fig. 15. Variation in hot-rolled texture with lithium content for (0001).
Fig. 16. Variation in contour location with lithium content.
As was mentioned earlier no single figure accurately represents the texture of a bulk specimen. However, now that it has been established how the texture varies from the surface to the centre of both high and low Li content alloys in both the hot and cold-rolled condition an "average" texture may be defined for each specimen. It can be seen after examining Figures 5, 6, 11 and 12 that a pole figure of a surface 3/8 of the way between the surface and the centre of the specimen best describes the average texture for all conditions. Therefore in the following tests all pole figures will be for this 3/8 depth unless otherwise specified.

(3) **Effect of Alloy Content on Hot-Rolled Texture**

The hot-rolled texture of pure Mg approaches the "ideal" texture for an hexagonal metal (i.e. all the basal planes aligned parallel to the sheet surface). To study the effect of Li on this near-ideal texture pole figures were prepared for 0.5, 2.5, 6 and 12.4 at. % Li alloys in the hot-rolled condition. These are shown in Figure 15. The contour locations on Figure 15 is shown in the graphical form of Figure 16.

The most obvious feature of Figure 16 is the rapid loss in sharpness from pure Mg through the 2.5 at. % Li alloy. From 2.5 % to 12.4 at. % the rate of change is much less.

(4) **Effect of Cold Work on Hot-Rolled Texture**

The effect of cold working a hot-rolled alloy varies with the alloy studied. For pure Mg the relatively symmetrical pole figure of the hot-rolled material is slightly elongated in the positive and negative rolling directions with 17% cold work (Figures 17 and 18). In the case of
Fig. 17. Variation in hot-rolled texture of pure Mg with cold work for (0001).
Fig. 18. Variation in contour location with cold work for pure Mg.
Fig. 19. Variation in hot-rolled texture of 12.4 at.% Li alloy with cold work for (0001).
Fig. 20. Variation in location of 100% peaks with cold work for 12.4 at.% Li alloy.
Fig. 21. Effect of cold work on hot-rolled texture for 0.5 and 2.5 at.% Li alloy for (0001).
Fig. 22. Effect of cold work on hot-rolled texture for 6 at.% Li alloy for (0001).
Fig. 23. Effect of depth on texture of 16% cold rolled 12.4 at.% Li alloy for (0001).
the highest alloy examined, the 12.4 at. % Li, only 16% cold work causes a complete split in the 100% intensity contour (Figure 19). The effect of cold work on the degree of splitting is shown in Figure 20. It is also significant to note that although the pole figure becomes more diffuse in the rolling directions with cold work the intensity for the 100% peak increases with total cold work (Figure 20). This results from the corresponding sharpening of the texture in the transverse rolling directions. This means the number of basal planes aligned in the transverse directions has been reduced by the cold working.

Figures 21 and 22 show the change in the hot-rolled texture of the 0.5, 2.5 and 6 at. % Li alloys. That the splitting in the 100% contour depends upon alloy content, cold work and depth of the section examined below the rolled surface is shown in Figures 19, 22 and 23. A complete split occurs in the 12.4 at. % Li alloy after 16% cold work but does not appear in the 6 at. % Li alloy until 60% cold work. After 16% cold work the 12.4 at. % Li alloy shows a near "ideal" texture at the surface and 0.002" below the surface but shows a split texture at 3/8 depth (Figure 23).

(A)2. Discussion of Texture Results

(1) Hot-Rolled Texture of Mg and Mg-Li Alloys

The following results need to be explained in light of the stress system and deformation systems active during the hot rolling of Mg-Li alloys:

a) the addition of Li to Mg causes a loss in sharpness of the (0001) texture (Figure 15).
b) the (0001) texture loses sharpness with depth below the as-rolled surface (Figures 3 and 4). In both the 0.5 at. % Li and the 12.4 at. % Li alloys there is a pronounced elongation in the rolling direction but appreciably less change in the transverse direction (Figures 5 and 6).

c) there is no obvious indication of <1120> directional texture (Figures 7 and 8).

1) Deformation Systems Active During Hot Rolling

The deformation systems active during hot rolling will be examined first.

In pure Mg tested at 400°C the CRSS for both first-order prismatic and first-order pyramidal slip is 200 gr./mm². Comparable CRSS values for basal slip are 60 gr/mm² (9) and 43 gr/mm² (4).

This difference in the CRSS between basal and nonbasal slip indicates that although the majority of the deformation occurring during hot rolling will be by basal slip some prismatic and pyramidal slip will occur in grains unfavourably oriented. The slip that occurs on the prismatic and pyramidal planes will of course reduce the sharpness of the (0001) basal plane texture.

As shown in Table 4, the increase in Li content increases the CRSS for both basal and nonbasal slip at 400°C. The values for the 15 at. % Li alloy of Yoshinaga et al (9) show that prismatic and pyramidal systems have the same CRSS at 400°C. Their values are much lower than those of Ahmadieh et al (10).

Using the values reported by Yoshinaga et al (9) for basal and nonbasal slip it becomes apparent that the relative amount of nonbasal
### TABLE 4

Critical Resolved Shear Stress

**BASAL SLIP (grams/mm²)**

<table>
<thead>
<tr>
<th>Alloy</th>
<th>400°C</th>
<th>Room Temp.</th>
</tr>
</thead>
<tbody>
<tr>
<td>Mg</td>
<td>43 (4)</td>
<td>50 (4)</td>
</tr>
<tr>
<td>Mg</td>
<td>60</td>
<td>75</td>
</tr>
<tr>
<td>0.4 at. % Li</td>
<td>75</td>
<td>90</td>
</tr>
<tr>
<td>3.7</td>
<td>150 (9)</td>
<td>230 (9)</td>
</tr>
<tr>
<td>6.6</td>
<td>230</td>
<td>335</td>
</tr>
<tr>
<td>15</td>
<td>260</td>
<td>460</td>
</tr>
<tr>
<td>12.5</td>
<td>420 (30)</td>
<td>560 (30)</td>
</tr>
</tbody>
</table>

**NONBASAL SLIP (grams/mm²)**

<table>
<thead>
<tr>
<th>400°C</th>
<th>System</th>
<th>Room Temp.</th>
<th>System</th>
</tr>
</thead>
<tbody>
<tr>
<td>Mg</td>
<td>200</td>
<td>prismatic</td>
<td>3,800</td>
</tr>
<tr>
<td>Mg</td>
<td>200 (9)</td>
<td>pyramidal</td>
<td>3,500</td>
</tr>
<tr>
<td>15 at. % Li</td>
<td>400</td>
<td>prismatic</td>
<td>------</td>
</tr>
<tr>
<td>15</td>
<td>400</td>
<td>pyramidal</td>
<td>------</td>
</tr>
<tr>
<td>Mg</td>
<td>102</td>
<td>prismatic</td>
<td>6,940</td>
</tr>
<tr>
<td>7.9 at. % Li</td>
<td>820 (10)</td>
<td>prismatic</td>
<td>3,580</td>
</tr>
<tr>
<td>12.9</td>
<td>2,300*</td>
<td>prismatic</td>
<td>2,860</td>
</tr>
<tr>
<td>15.9</td>
<td>2,400*</td>
<td>prismatic</td>
<td>2,660</td>
</tr>
</tbody>
</table>

* Extrapolated from 130°C.
slip occurring during rolling at 400°C will increase with increasing Li content since the ratio of the CRSS for basal to nonbasal slip decreases with increasing Li content. If the value of CRSS for basal slip of Quimby, Mote and Dorn (30) is correct then the CRSS for basal and nonbasal slip must be almost equal at 400°C for high Li alloys.

The significant increase in the relative amount of nonbasal slip with increasing Li content explains the corresponding loss of sharpness in the (0001) texture since the active prismatic and pyramidal planes will rotate towards the sheet surface under the influence of both the tensile and compressive forces. This will rotate the corresponding basal planes away from the sheet surface.

ii) **Stress System During Hot Rolling**

The second factor to consider in the development of textures during rolling is the stress system operative. This is generally considered as being biaxial: consisting of a tensile stress in the rolling direction and a compressive stress normal to the rolling direction. Stress in the transverse direction is considered to be negligible because of the small transverse spread observed during rolling.

Although the stress system is biaxial throughout the full cross-section of the material the resulting deformation varies between the surface and the centre. Crane and Alexander (17) have shown that when "sticking friction" occurs between the rolls and the material, as it does in the hot rolling of Mg, the surface metal is drawn ahead of the bulk as it enters the mill because the rolls are moving at a higher speed than the material. However, on leaving the mill the surface material is held back. This produces the overall deformation shown in Figure 24.
Fig. 24. Deformation during rolling.
This variation in the deformation from the surface to the centre is the reason for the observed change in texture with depth. The compressive deformation of the surface material will produce a symmetrical pole figure since deformation planes of all orientations are activated to varying degrees, while the combined compressive and tensile deformation in the bulk of the material will produce a pole figure that is elongated in the rolling direction as a result of the tensile deformation. This elongation in the rolling direction results from the greater effect of the tensile stress on the nonbasal planes associated with those basal planes oriented in the rolling direction. Also this effect is most pronounced, as will be shown in a later section, on those basal planes having the least spread in texture from the ideal.

The lack of any $<1\bar{2}0>$ directional texture is not surprising since up to three separate systems are operating at once $\{10\bar{1}0\} <1\bar{2}0>$; $\{10\bar{1}1\} <1\bar{2}0>$; and $(0001) <1\bar{2}0>$, and since it only takes a small angular scatter in the directional texture to obscure its presence.

(2) Cold-Rolled Texture of Mg and Mg-Li Alloys

The change in the hot-rolled texture of Mg and Mg-Li alloys with cold rolling is outlined below:

a) in pure Mg and 0.5 at. % Li the cold rolling has little effect on the texture other than to increase the relative intensity of the 100% reflection (Figures 17 and 21).

b) in the 2.5, 6 and 12.4 at. % Li alloys the cold rolling causes a split in the 100% intensity contour (Figures 19, 21 and 22), and a general elongation of the basal contours in the rolling direction. The amount of cold work required to cause the split decreases with increasing Li content.
c) the pure Mg and all the alloys show a variation in texture with depth below the as-rolled surface.

d) both low and high Li alloys show a directional texture at the surface but show a random $\langle 11\bar{2}0 \rangle$ texture below the surface.

The changes observed in the hot-rolled texture with cold rolling imply a definite change in the deformation mechanisms operative during cold rolling. The factors which affect the deformation mechanisms are:

a) the ratio of the CRSS for basal and nonbasal slip

b) the stress system present during rolling

c) the degree of constraint imposed on the grains

i) Deformation Systems Active During Cold Rolling

The ratio of the CRSS for basal and prismatic slip in both pure Mg and 15 at. % Li at 400°C and room temperature is shown in Table 5.

<table>
<thead>
<tr>
<th>Alloy</th>
<th>400°C CRSS Prismatic/CRSS Basal</th>
<th>Room Temperature CRSS Prismatic/CRSS Basal</th>
</tr>
</thead>
<tbody>
<tr>
<td>Pure Mg</td>
<td>$\frac{200}{60} = 3.3$ (9)</td>
<td>$\frac{3,800}{75} = 50.7$ (9)</td>
</tr>
<tr>
<td>15 at % Li</td>
<td>$\frac{2,400}{260} = 9.2$ (10)</td>
<td>$\frac{2,660}{460} = 5.8$ (10)</td>
</tr>
</tbody>
</table>

It is quite apparent that little if any prismatic slip can be expected in pure Mg at room temperature while the possibility of prismatic slip in the 15 at. % Li alloy is quite high. Metallographic examinations made on hot-rolled alloys tested in tension at room
temperature confirmed that for Li contents of 2.5 at. % and higher nonbasal slip definitely occurs. Details of this test are given in a later section.

In the cold rolling of both pure Mg and Mg + 0.5 at. % Li the cold work does not have a pronounced effect on the texture (Figures 17 and 21). The 0.5% alloy shows no appreciable change in contour location but does show an increase in intensity of the 100% value with cold work. The pure Mg shows an even greater increase in the 100% value but also shows some spread in texture with cold work. Since mainly basal slip occurs in these materials, and since basal slip will produce a sharpening rather than a weakening of the texture, it seems likely that the more diffuse nature of the 17% cold-rolled Mg specimen may result from the section being taken slightly deeper than anticipated. As shown in Figure 9 the effect of depth on the texture is very marked.

The cold rolling of 2.5, 6 and 12.4 at. % Li alloys causes a marked spread of the basal texture in the rolling direction and a splitting of the 100% contour in the 6 and 12.4 at. % alloys. This results from the nonbasal slip occurring. The ratio of the CRSS for basal and nonbasal slip (Table 5) shows that nonbasal slip can occur in the higher Li alloys.

Because of the 90° angle between the basal and prism system they both subtend the same angle with the 45° maximum shear plane. Assuming basal slip occurs first because of its lower CRSS the grain will rotate to align the basal plane parallel to the sheet surface. As rotation occurs the constraint imposed on the grain by the surrounding matrix causes strain hardening on the operative system. Once the CRSS
on the operative basal system exceeds the value for prism slip the prism system will become operative. At this point major grain rotation will stop and, as the prism and basal systems are both now intermittently operative, the grain orientation will fluctuate around this equilibrium value. Further cold work should increase the concentration of poles around this equilibrium value.

The value of the equilibrium angle for the basal planes relative to the sheet surface should act as a qualitative measure of the difference in the CRSS for basal and prism slip. In pure Mg the difference is very high and the basal planes align themselves parallel to the surface. In the case of 6 and 12.4 at. % Li alloys comparing the 60% cold work figures the peak locations are 10° and 14° (Figure 22) and 11° and 19° (Figure 20). The difference in the CRSS for basal and prism slip decreases with increasing Li content.

ii) Stress System During Cold Rolling

The stress system present in cold rolling should be the same from the surface to the centre and will consist of a tensile stress in the rolling direction and a compressive stress normal to the sheet surface. According to Crane the total deformation occurring varies only slightly from the surface to the centre, being highest at the surface and lowest at the centre.

iii) Grain Constraint During Cold Rolling

An important factor in the texture produced during cold rolling is the variation in constraints on grains at the surface and those in the bulk of the material. As a result of the slippage between the rolls and the sheet the grains at the surface have one relatively
free surface and can rotate more freely during deformation than can those at the centre. Because of the greater constraints imposed on the grains in the bulk of the material, the effective stress acting on the second order deformation systems $\{10\overline{1}0\} \langle 1\overline{1}20 \rangle$ can build up to the CRSS value of that system very rapidly during the deformation on the first-order system (0001) $\langle 1\overline{1}20 \rangle$. Since the average grain size of the alloys examined is 50 microns ($\mu$) the surface layer of grains extends 0.002 in. into the sheet and represents the entire depth examined by the surface pole-figure.

In the pure Mg and the 0.5 at. % Li alloy the variation of texture with depth is a direct result of the difference in the grain constraint between the surface and the centre. Since the surface grains are relatively free to rotate during deformation and since basal slip is the only system operative to any extent, the surface material should show a sharper (0001) texture than the interior material. This occurs as is shown in Figure 9. The relative intensities for the 100% reflection are also higher for the surface pole-figure than for those of the interior.

The effect of the tensile component of the rolling stress is manifest in the sharp $\langle 1\overline{1}20 \rangle$ directional texture present at the surface of the 0.5 at. % Li alloy and to a lesser extent the 12.4 at. % Li alloy (Figures 13 and 14). The reduced constraint on the surface grains is again apparent since the pole-figures for the interior, where grain rotation is inhibited, show a random directional texture. The slight difference between the two interior (0001) pole-figures (Figure 9) is probably due to the difference in the deformation with
depth. As depicted in Cranes' sketch of the deformation of a vertical plane in cold rolling (Figure 24), the midway to centre material will receive somewhat more deformation than the centre material and will therefore show a slightly sharper (0001) texture.

The effect of depth on the (0001) texture of the 12.4 at. % Li alloy is shown in both Figure 10, for 60% cold work, and Figure 23, for 16% cold work. As already observed there is a marked loss in texture with depth. There is also a change from a definite directional texture at the surface to a random texture below the surface (Figure 14).

These results show that the surface grains, being subjected to less constraint, deform mainly by basal slip. This produces less spreading of the (0001) texture than occurs below the surface where appreciably more prismatic slip occurs because the grain constraints increase the effective stress acting on the nonbasal systems. This argument also holds for the (1120) directional texture which is present at the surface where grain rotation can occur more freely and not present below the surface where grain rotation is inhibited and where other systems are also active.

(b) Tensile Properties of Mg-Li Alloys

The tensile properties of Mg and four Mg-Li alloys were determined by the test methods described in the Experimental Section.

(1) Preparation of Tensile Specimens

As mentioned in the Experimental Section B 4 several tests were run to compare the effects of stamping and spark machining of the tensile specimens. Metallographic examination of the stamped and spark machined edges showed that after 0.001 in. of metal had been removed by
pickling there was no evidence of any deformation from either method of preparation. Since all tensile specimens were to be pickled prior to testing no problem was anticipated from using stamped specimens. However, comparison of tensile results run on identical material prepared by the two techniques showed in the case of as-hot-rolled commercially pure Mg prepared by stamping and pickling the yield strength was increased somewhat over that of spark machined and pickled specimens. Since this introduced some doubt into the effect of the stamping on the properties all specimens were subsequently prepared by spark machining.

(2) Effect of Li on Tensile Properties of Hot-Rolled Mg

The effect of additions of Li to pure Mg is shown in Figures 25 and 26. The results shown in these figures and presented in Table 6 are for an average grain size of 50 microns. As shown in Table 6 the true grain size varied from 20 to 80 microns. The values plotted in Figures 25 and 26 are corrected to the 50 micron value using data on grain size vs strength for Mg-Li alloys presented by Hauser et al. (8). Although grain size control by annealing treatments would have allowed the production of a constant grain size from alloy to alloy it was felt that this would alter the normal as-hot-rolled conditions being studied and therefore was not used.

A comparison of the yield strength, ultimate strength, and ductility of the alloys tested by Hauser et al (8) (Figure 27), Yoshinaga et al (9) (Figure 28) and the present work, is given in Table 7 and Figures 29 and 30. The analysis given for the alloys (see Table 2) shows that the material used by Hauser et al was of a higher purity than that used in the present investigation particularly with regards
Fig. 25. Uniform elongation and 0.2% offset yield strength vs. at.% Li for hot-rolled alloys.
Fig. 26. Ultimate strength vs. at.% Li for hot-rolled alloys.
Fig. 27. Tensile properties vs. at.% Li for hot-rolled Mg-Li alloys (Ref. 8).
Fig. 28. Tensile properties vs. at.% Li for hot-rolled Mg-Li alloys (Ref. 9)
Fig. 29. Comparison of yield and ultimate strengths for hot-rolled Mg-Li alloys.
Fig. 30. Comparison of ductilities for hot-rolled Mg-Li alloys.
to the Ca content (0.01% to 0.07% respectively). A detailed comparison is given in Table 2. The material used by Yoshinaga and Horiuchi was reported as 99.99% Mg and 99% (Na + K 0.03%) Li with no analysis of the impurity level after alloying being given.

The present results agree very well with those of Yoshinaga and Horiuchi for all the tensile properties reported. The curve for the ultimate strength vs at. % Li from Yoshinaga's results is shown dotted since the values plotted result from an extrapolation to the indicated fracture strain for all strains beyond 10%. However, although the absolute values may be slightly low the shape of the curve is similar.

Agreement between the present results and those of Hauser et al is not as good. The main disagreement occurs in the ultimate strength curve. Here Hauser shows a distinct minimum at the 4.4 at. % Li alloy followed by a slow but steady increase for higher Li alloys. Both Yoshinaga's results and the present work show a continual decrease in ultimate strength with increasing Li content from approximately 0.5 at. % Li. The reason for the disagreement is not immediately apparent especially since the impurity level difference between Hauser's material and that of this report is greatest for the higher Li alloys (i.e. the present material having the higher impurity level but showing lower strengths).

The yield strengths plotted from Hauser's results may not be completely accurate because of the extreme crowding on his graph. Generally the agreement in yield strengths is fairly good, except Hauser's values level off from 7.6 at. % Li to the 14.5 at. % Li alloy.
The correlation of mechanical properties and deformation mechanisms operative must be based on the results discussed in the introduction. The single crystal results of most interest are the reported increase in the CRSS for basal slip with increasing Li content (9) and the decrease in CRSS for prismatic slip with increasing Li content (for low, i.e. ambient, temperatures) (10). As noted by Yoshinaga and Horiuchi (9) the fact that the 0.4 at. % Li alloy shows no increase in ductility over that of pure Mg implies that only basal slip is operative in this alloy. Therefore, the increase in strength shown is entirely due to basal slip hardening. Metallographic examination of a 0.5 at. % Li longitudinal tensile specimen tested to fracture in the present series of tests showed no evidence of nonbasal slip thereby strongly supporting this postulate.

Once the alloy level reaches 2.5 at. % Li the ductility has increased to such an extent that appreciable prismatic slip is occurring. By 6 at. % the maximum ductility has been obtained. Therefore for alloys of 2.5 at. % Li and higher, consideration must be given to both basal and prismatic slip being operative during deformation. Confirmation of the postulate that two slip systems are active was obtained by metallographic examination of a 2.5 at. % and 12.4 at % Li alloy after tensile testing to failure. The 2.5 % alloy had two slip systems operative in regions adjacent to the fracture while the 12.4 at. % alloy had two slip systems operative in regions well removed from the fracture. Since X-ray diffraction patterns were not run on the grains showing duplex slip it can not be insured that the second system operating was prismatic slip. The literature gives no evidence of
extensive pyramidal slip occurring in Mg-Li alloys at room temperature, however, so it is probably safe to assume that the second set is the prismatic system.

This means that, since the CRSS for prismatic slip decreases with increasing Li content, the stress necessary to deform the tensile specimens should decrease from that Li content at which prismatic slip becomes significant. In light of the preceding discussion this is somewhere between 0.5 and 2.5 at. % Li. The Li added up to this level causes solid solution hardening but Li added beyond this point reduces the solid solution hardening effect by introducing a second operative slip system. The higher the Li content the lower the CRSS for the second system and the lower the tensile strength of the material. Naturally since the second system offers additional modes of deformation the ductility will increase. The loss in sharpness of the basal plane texture observed with the increase in the Li content will result in lower yield and ultimate strengths because of the improved Schmidt Factor for basal slip.

(3) **Effect of Cold Work on Tensile Properties of Hot-Rolled Mg-Li Alloys**

Four Mg-Li alloys were cold rolled to varying reductions and their tensile properties were determined. Figures 31 to 37 show the effect of cold work and Li content on the following quantities:

i) 0.2% offset yield strength ($\sigma_{YS}$)

ii) ultimate strength ($\sigma'$)

iii) the ratio of stress change to strain change for comparable degrees of cold work
Fig. 31. 0.2% offset yield strength vs. % cold work for Mg-Li alloys.
Fig. 32. Ultimate strength vs. % cold work for Mg-Li alloys.
Fig. 33. Ratio $\Delta \sigma / \Delta \epsilon$ for three increments of cold work.
iv) the ratio of transverse to longitudinal tensile and yield strengths

v) uniform elongation \( (\varepsilon_u) \)

vi) the ratio of elongation to fracture to uniform elongation \( (\varepsilon_f/\varepsilon_u) \)

It should be noted that the hot-rolled strengths (0% cold work) noted for each of the above cases is that actually measured and contains no correction factor for grain size effect since the cold-rolled values cannot have the same correction factor applied.

In examining the results shown in Figures 31 to 37 the following factors must be considered:

i) the difference in the effect of cold work on the different alloy levels

ii) the change in the rate of strain hardening with increasing cold work

iii) the difference in the effect of cold work on the longitudinal and transverse properties

To explain the above, the effect of cold work on the two major deformation systems operative must be considered.

The increase in both yield strength and ultimate strength with cold work represents the usual hardening of the active deformation systems during deformation. In the alloys studied it has been observed that basal slip is the prime deformation mechanism in both Mg and Mg + 0.5 at. % Li while prismatic slip becomes increasingly important with increasing Li content.

Considering the yield strength values the 0.5 at. % Li
alloy and pure Mg show a lower rate of strain hardening during cold rolling than do the higher Li alloys. This shows in the slope of the curves in Figure 31 and in the curves of Figure 33. The three graphs of Figure 33 show the change in strength resulting from specific change in strain introduced by cold rolling. The three strain levels considered are 0 to 16%, 16 to 30% and 30 to 60% reduction in area. The reason for the difference in the rate of strain hardening with the Li content is the difference in the rate of strain hardening on the two major slip systems. Yoshinaga and Horiuchi (6) reported that the rate of strain hardening for nonbasal slip was very high while that for basal slip was much lower in the single crystal tests they ran. Therefore the greater the relative amount of nonbasal slip the greater the rate of strain hardening as was observed. Similar arguments hold for the ultimate strength results (Figure 32).

The ratio of the transverse to longitudinal tensile properties (Figures 34 and 35) show that

i) the transverse strengths exceed the longitudinal strength in all cases.

ii) the ratio $\frac{\sigma_T}{\sigma_L}$ decreases with increasing Li content for comparable cold work (i.e. whether numerically equal reductions Figure 34, or terminal cold work values, Figure 35, are considered).

iii) the $\frac{\sigma_T}{\sigma_L}$ ratio increases with increasing cold work for any one Li content.
Fig. 34. Ratio transverse/longitudinal tensile properties vs. % cold work.
Fig. 35. Ratio transverse/longitudinal tensile properties vs. at. % Li for "Terminal cold work condition".
The explanation of the transverse strength always exceeding the longitudinal strength lies in the texture of the alloys. All alloys showed an elongation of the pole-figures in the rolling direction. This means that the number of basal planes with orientations removed from the "ideal" sheet texture is greater in the longitudinal direction than in the transverse direction which in turn means a lower CRSS for basal slip in the longitudinal direction.

The decrease in $\sigma^y$ for increasing Li content is explained by the introduction of a second slip system (i.e. prismatic) in the higher Li alloys. This decreases the effect of texture on the transverse specimens.

The increase in $\sigma^t/\sigma^l$ with increasing cold work for a particular Li content indicates a higher rate of cold working in the transverse direction than in the longitudinal direction. This may be a result of the increased amount of prismatic slip required in the transverse direction relative to the longitudinal direction since the texture is such that basal slip is inhibited in the transverse direction and prismatic slip must occur to preserve grain continuity. Since the prismatic system work hardens more rapidly than the basal system the transverse direction will show a higher hardening rate than the longitudinal direction with increasing cold work as was observed.

The values shown in Figure 36 indicate that the ductility increases with increasing Li content and decreases with increasing cold work. The first of these is expected as a result of the increasing amount of prismatic slip while the second is a normal hardening effect. In all but the 6 at. % Li case the transverse ductility exceeds the
Fig. 36. % Uniform elongation vs. % cold work for Mg-Li alloys.
longitudinal ductility in the cold-worked condition but is less in the hot-rolled condition.

In the hot-rolled alloys the (0001) texture shows a slight spread in the rolling direction which, as explained above, means more basal planes are aligned for slip than in the transverse direction. This produces higher longitudinal ductility. This will be important in both the low Li alloys where only basal slip occurs and in the high Li alloys where basal slip still contributes a significant percentage of the deformation. This is particularly true when it is remembered that the prismatic system hardens more rapidly than the basal system. No reason is available for the higher transverse ductility after cold work nor for the anomalous results for the 6 at. % Li alloy.

The elongation to fracture consists of the uniform elongation plus the elongation during necking. In the low Li alloys the ratio of elongation to fracture over the uniform elongation is unity (Figure 37) and indicates no necking occurs. Since necking of a test piece requires prismatic slip and since basal slip is the prime deformation system in the low Li alloys this result is as expected.

In the higher Li content alloys, the amount of prismatic slip increases but, because of the spread of the (0001) texture in the rolling directions, the amount of basal slip also increases. This means that in the hot-rolled case the uniform elongation will increase but the $\varepsilon_f/\varepsilon_u$ ratio will remain near 1.0 for both longitudinal and transverse directions. Once the higher Li alloys are cold worked a further spreading of the basal texture keeps the $\varepsilon_f/\varepsilon_u$ ratio for longitudinal specimens at 1.0 by increasing the relative ease of basal slip over prismatic slip. Especially with the more rapid
Fig. 37. Ratio $\epsilon_f/\epsilon_u$ vs. $\%$ cold work for Mg-Li alloys.
hardening of the prismatic system. In the transverse direction the basal texture remains unchanged or is sharpened. Here enough prismatic slip will occur to cause the $\varepsilon_f/\varepsilon_u$ ratio to increase with increasing cold work (see the 6 at. % Li curves).

The reason for the shape of the $\varepsilon_f/\varepsilon_u$ curves for the 12.4 at. % Li alloy is the large splitting of the basal plane texture. The highest concentration of basal planes take orientations of up to $20^\circ$ from "ideal". This splitting begins with the first cold work and passes through a maximum around 30% cold work. With the main concentration of basal planes as much as $20^\circ$ from the sheet surface, and with prismatic slip common, the amount of necking observed should increase with increasing spread in basal texture. Therefore, the $\varepsilon_f/\varepsilon_u$ curve should have the same shape as the location of the basal plane maximum as is observed (compare Figures 20 and 37).

(C) The Effect of Texture on the Tensile Properties of Mg-Li Alloys

The effect of texture on the tensile properties can be major as has been discussed. The two methods described for correlating texture and tensile properties that were thought to be applicable to the present work are those of Couling (28) and Avery et al (18).

(1) Correlation of Texture and Strength After Couling

The method of Couling for predicting the ratio of transverse to longitudinal properties consists of combining the Schmid Factor for any crystal orientation relative to the tensile test direction, $\sin \alpha \cos \alpha$, and the density factor for that orientation, $I_0 \sin \Theta$. The density factor consists of the relative intensity of the X-ray reflection for a particular orientation $I_0$ times the population factor for that orientation $\sin \Theta$. 

The general assumptions that must be made to use this method are:

a) the material is single-phase, strain-free, and equiaxed in grain structure. There is no banding. The pole-figure is representative of the entire specimen thickness.

b) there is no preferred alignment of the \( \langle 11\overline{2}0 \rangle \) slip direction.

c) for small strains (sufficient to reach the 0.2% offset strain) the only deformation mechanism active to an appreciable degree is basal slip.

d) slip obeys a resolved shear stress law, i.e. if the basal pole of a grain makes an angle \( \alpha \) with the tensile axis the relative ease of slip for that grain is simply \( \sin \alpha \cos \alpha \)

e) the tensile yield strength for the entire specimen is inversely proportional to the collective ease of slip for all the grains.

An example of this method follows. The specimen used was the 0.5 at. % Li alloy in the hot-rolled and 16% cold-rolled conditions. Graphs were prepared for the variation of intensity with angle for the \( \langle 0001 \rangle \) plane reflections on the pole figures using the rolling direction, transverse direction and 30° and 60° diameters between the rolling direction and the transverse direction (see Figure 38). These graphs gave the intensity factor, \( I_\theta \), used by Couling. A table of Schmid Factors times the population density factor (\( \sin \theta \)), was prepared and the complete \( I_\theta \sin \theta \cos \alpha \sin \alpha \) values were calculated for each diameter and for tensile loads in both the transverse and longitudinal directions.
Fig. 38. a) Directions used for Intensity vs. Angle Determination.

Fig. 38 b) Intensity vs. Angle for (0001) Planes.
Summing the values for each diameter and then summing the diameter totals for both longitudinal and transverse directions gives figures for comparative ease of slip in the two directions. The inverse ratio of the sum for the transverse over the sum for the longitudinal should equal the ratio of the transverse to longitudinal yield strengths if the texture is the only factor affecting the results. If only basal slip is occurring during the complete deformation process then the ultimate strength ratio should also have the same value.

The results of measured values and those obtained from Couling's method are given in Table 9. The results for the cold-worked material are in fair agreement while those for the hot-rolled material show no agreement at all. It is believed that the good agreement in the ultimate strength ratio for the 16% cold-rolled material is fortuitous since the yield strength ratio does not show the same agreement. The yield strength should agree if either does since only basal slip is expected for this amount of strain whereas some nonbasal slip and twinning should occur before the ultimate strain is reached.

Consideration of Couling's factor for the number of planes having a particular orientation led the author to propose a modified population factor which is based on $I_0$ only. Since the $I_0$ values on the pole-figure are an accurate measure of the number of poles, relative to the 100% value, having any particular orientation the $\sin \theta$ factor should not be used. Therefore the factor used by the author was $I_0 \sin \alpha \cos \alpha$. The results of this modified factor are also given in Table 9. Here the best agreement is with the yield strength values for both the hot-rolled and cold-rolled specimens. The results for the
ultimate strength is not as good. Since both twinning and some second order slip must occur between the yield and ultimate strains the lack of agreement in the ultimate strength ratios is understandable, while the relatively good agreement in the yield strength ratios is thought to be significant.

(2) Correlation of Texture and Strength After Avery and Hosford

The observations of Avery, Hosford, and Backofen (18) that the amount of change in the width and thickness of a tensile specimen during testing depended on the basal pole-figure spread led them to the following result. Since the orientation of the prism planes, assumed as the second deformation system active in Mg alloys at room temperature, is fixed relative to the basal planes a relationship between the Schmid Factor for each set of planes can be derived. This relationship is given below with the stereographic representation of the slip elements in a crystal.

In a given (isolated) grain

\[
\begin{align*}
\tilde{\tau}_{(000)\langle12\bar{1}0\rangle} &= m_{(000)} \cdot \sigma_{(\text{applied})} \\
\tilde{\tau}_{(000)\langle1\bar{2}10\rangle} &= m_{\langle\overline{0}0\overline{1}\rangle} \cdot \sigma_{(\text{applied})}
\end{align*}
\]

where the resolving factors are

\[
\begin{align*}
m_{(000)} &= \cos \Theta \sin \Theta \cos \alpha \\
m_{\langle\overline{0}0\overline{1}\rangle} &= \sin^2 \Theta \cos(30 + \alpha) \cos(60 - \alpha)
\end{align*}
\]

\(\Theta\) is the angle between the tensile axis and (0001), and \(\alpha\) is the angle between the great circle through the (0001) and the tensile axis and the great circle through the (0001) and the \(\langle1\bar{2}10\rangle\) slip direction nearest the tensile axis (see Figure).
The ratio of the shear stresses on the prism and basal planes is then \[
\frac{m_{[1010]}}{m_{[0001]}} = \tan \Theta \left[ \frac{\cos(30+\alpha) \cos(60-\alpha)}{\cos \alpha} \right]
\]

Since \(0 < \alpha < 30\) for the most highly stressed slip systems, the value of \(f(\alpha)\) can vary only between 0.433 and 0.524 with a mean about 0.50 and gives a curve for the dependence of \(\frac{m_{[1010]}}{m_{[0001]}}\) on the basal-pole spread, \((90-\Theta)\), as shown.

\[
\frac{m_{[1010]}}{m_{[0001]}} = \tan \Theta \cdot f(\alpha)
\]
This curve shows that as the (0001) spread increases the possibility of prism slip decreases, assuming that no change occurs in the ratio of the CRSS for basal and prism slip.

However, since the prismatic and basal slip systems appear to harden at different rates due to cold working, and since no CRSS values are available for cold-worked material, it does not appear possible to correlate the split in texture resulting from cold working and Avery's graph of $\frac{m_{[001]}^{[001]}}{m_{(000)}^{(000)}}$ vs (0001) spread. If CRSS values were available some measure of the relative amount of basal and prism slip could be obtained.
IV. CONCLUSIONS

(1) It has been shown that a pole-figure only represents the texture for a thin surface layer of metal. Unless a series of pole-figures is prepared representing various depths, or an average pole-figure is drawn based on this series, a completely misleading picture will result of the texture of the material.

(2) The addition of Li to Mg has a marked effect on the rolling texture. In all cases the (0001) texture becomes more diffuse or less ideal with increasing Li content. No indication of a definite $<1\bar{1}20>$ directional texture occurred in the hot-rolled alloys.

(3) The change in rolling texture with Li content was explained successfully on the basis of deformation systems active during rolling.

(4) Cold working of the alloys reduced the sharpness of the (0001) "ideal" texture. In the low Li alloys the effect was small while in the higher Li content alloys a pronounced split in the texture occurred.

(5) A definite $<1\bar{1}20>$ directional texture was observed at the surface of the 0.5 at. % Li alloy after cold rolling.

(6) Again the change in texture was explained successfully on the basis of deformation systems active during cold rolling.

(7) The addition of Li to Mg causes a sharp increase in the 0.2% offset yield and ultimate tensile strengths for hot-rolled alloys up to approximately 0.5 at. % Li. Beyond this level the strengths drop steadily to 12.4 at. % Li.

(8) The ductility, as indicated by uniform and total elongation, remains relatively constant to 0.5 at. % Li and then increases markedly.
(9) The variation of tensile properties with Li content agreed very well with those of Yoshinaga and Horiuchi (9) but showed some discrepancy to those of Dorn et al (8).

(10) The variation in tensile properties with Li content was explained on the basis of the deformation systems active (as indicated by the relative amount of basal and prismatic slip).

(11) The effect of cold work on such tensile properties as yield and ultimate strengths, total and uniform elongation, and rate of work hardening have been explained on the basis of active deformation systems.

(12) An attempt at predicting the transverse to longitudinal tensile strength ratios on the basis of the observed texture was successful for the yield strength in the 0.5 at. % Li alloy, both as hot rolled and after 16% cold work, but was unsuccessful for the ultimate strength in both conditions.

V. AREAS FOR FUTURE WORK

(1) The effect of annealing on the properties of cold-worked Mg-Li alloys requires a great amount of work. A definite answer as to whether or not there is a strengthening effect on annealing, and the cause of the yield point observed in the 12.4 at. % Li alloy would give a great deal of indirect evidence as to the slip systems active in the cold-worked alloys.
VI. BIBLIOGRAPHY


14. E.A. Calnan, C.J.B. Clews "The Development of Deformation Textures in Metals"
   Part II Body Centred Cubic, Phil Mag. 42, p. 616 (1951)


### TABLE 2

Chemical Analysis of Alloys (weight %)

<table>
<thead>
<tr>
<th>Alloy No.</th>
<th>Al</th>
<th>Zn</th>
<th>Cu</th>
<th>Fe</th>
<th>Mn</th>
<th>Ni</th>
<th>Pb</th>
<th>Si</th>
<th>Sn</th>
<th>Li</th>
<th>Ca</th>
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<td>&lt;0.001</td>
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<td>&lt;0.01</td>
<td>&lt;0.01</td>
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<td>&quot;</td>
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<td>&quot;</td>
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<td>&quot;</td>
<td>&quot;</td>
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<td>0.73</td>
<td>0.02</td>
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<td>Hauser et al</td>
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<td>&lt;0.01</td>
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Analysis are spectrographic except Li which is by flame photometry.
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<tr>
<th>Alloy At. %</th>
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<th>0.5</th>
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<td>790/820</td>
<td>775/800</td>
<td>775/800</td>
<td>725/750</td>
</tr>
<tr>
<td>°F</td>
<td>825/850</td>
<td>790/820</td>
<td>775/800</td>
<td>775/800</td>
<td>725/750</td>
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<tr>
<td>% Reduction</td>
<td>Max.</td>
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<td>24</td>
<td>26</td>
<td>25</td>
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<tr>
<td>in Thickness</td>
<td>Min.</td>
<td>12</td>
<td>11</td>
<td>14</td>
<td>13</td>
</tr>
<tr>
<td>Per Pass</td>
<td>Aug</td>
<td>20</td>
<td>18</td>
<td>20</td>
<td>16</td>
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<td>17</td>
<td>30</td>
<td>60</td>
<td>60</td>
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<tr>
<td>Obtained</td>
<td>All reductions 2% pass or less</td>
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## TABLE 6
Mechanical Properties of Hot Rolled Mg-Li Alloys

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<th>Alloy at. % Li</th>
<th>Grain Size μ</th>
<th>d^{-1/2} mm</th>
<th>Ultimate Strength</th>
<th>Correction Factor</th>
<th>Corrected Ultimate</th>
<th>0.2% Yield Strength</th>
<th>Correction Factor</th>
<th>Corrected Yield</th>
<th>ε_f/ε_u</th>
<th>% ε_u</th>
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<td>3</td>
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<td>30,200</td>
<td>11,700</td>
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<td>13,600</td>
<td>1.12</td>
<td>5.0</td>
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<td></td>
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<td>T25,000</td>
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<td>30,000</td>
<td>16,100</td>
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<td>18,700</td>
<td>1.10</td>
<td>2.1</td>
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<td>0.5</td>
<td>83</td>
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<td>L24,400</td>
<td>1.08</td>
<td>L26,300</td>
<td>12,900</td>
<td>1.10</td>
<td>14,200</td>
<td>1.00</td>
<td>5.3</td>
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<td></td>
<td>T25,500</td>
<td></td>
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<td>6</td>
<td>L24,300</td>
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<td>L22,400</td>
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<td>8,200</td>
<td>1.04</td>
<td>19.3</td>
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### TABLE 7
Comparison of Results From Present Work, Dorn et al, and Yoshinaga and Horiuchi

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<thead>
<tr>
<th>Tensile Property</th>
<th>Alloy (at. % Li)</th>
<th>Ref. (9)</th>
<th>Ref. (8)</th>
<th>Ref. (9) (8)</th>
<th>Ref. (9) (8)</th>
<th>Ref. (8)</th>
<th>Ref. (8) (9)</th>
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</thead>
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<td>0.4; 0.5</td>
<td>2.5; 2.6</td>
<td>3.7; 4.4</td>
<td>6.0; 6.6; 7.6</td>
<td>10.4</td>
<td>12.4</td>
</tr>
<tr>
<td>Present</td>
<td></td>
<td>26,300</td>
<td>24,800</td>
<td>22,400</td>
<td>20,900</td>
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<tr>
<td>Ref. 8</td>
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<td>22,100</td>
<td>23,000</td>
<td>23,500</td>
<td>17,500</td>
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<td>Ref. 9</td>
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<td>19,500</td>
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<tr>
<td>Present</td>
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<td>14,200</td>
<td>12,700</td>
<td>9,900</td>
<td>7,900</td>
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<tr>
<td>Ref. 8</td>
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<td>13,500</td>
<td>8,000</td>
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<td>7,700</td>
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<td>Ref. 9</td>
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<td>9,750</td>
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<tr>
<td>Present</td>
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<td>5.3</td>
<td>10.0</td>
<td>20.3 (26.8)</td>
<td>20.7 (27.5)</td>
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<td>Ref. 8</td>
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<td>2.8</td>
<td>3.0</td>
<td>8.8 (9.0)</td>
<td>10.8 (11.5)</td>
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<td>27.5</td>
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<td>Ref. 9</td>
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<td>7.6</td>
<td>15.5</td>
<td>26.3</td>
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### TABLE 8

**Mechanical Properties After Cold Working**

<table>
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<tr>
<th>Alloy</th>
<th>Property</th>
<th>Hot Rolled</th>
<th>16% C.R.</th>
<th>30% C.R.</th>
<th>60% C.R.</th>
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<tbody>
<tr>
<td>Comm.</td>
<td>ω</td>
<td>L</td>
<td>11,700±150</td>
<td>13,100±200</td>
<td>Actually 17% C.R.</td>
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<tr>
<td>Purity</td>
<td>ω</td>
<td>T</td>
<td>16,100±300</td>
<td>18,500±100</td>
<td>Some microcracks in longitudinal sections (max. approx. 15% long.)</td>
</tr>
<tr>
<td>Mg</td>
<td>α</td>
<td>L</td>
<td>25,100±500</td>
<td>17,500±200</td>
<td>Some microcracks observed in longitudinal sections (max. approx. 15% long.)</td>
</tr>
<tr>
<td></td>
<td>α</td>
<td>T</td>
<td>25,000±500</td>
<td>27,800±100</td>
<td></td>
</tr>
<tr>
<td></td>
<td>ε&lt;sub&gt;α&lt;/sub&gt;</td>
<td>L</td>
<td>5.0±1.5</td>
<td>1.2±0.1</td>
<td></td>
</tr>
<tr>
<td></td>
<td>ε&lt;sub&gt;α&lt;/sub&gt;</td>
<td>T</td>
<td>2.1±0.2</td>
<td>3.6±0.1</td>
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<tr>
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<td>ε&lt;sub&gt;ε&lt;/sub&gt;</td>
<td>L</td>
<td>1.1±0.0</td>
<td>1.2±0.1</td>
<td></td>
</tr>
<tr>
<td></td>
<td>ε&lt;sub&gt;ε&lt;/sub&gt;</td>
<td>T</td>
<td>1.0</td>
<td>3.6±0.1</td>
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<td>0.5</td>
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<tr>
<td>At.% Li</td>
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<td>17,800±400</td>
<td>17,000±200</td>
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<td>ω</td>
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<td>24,400±50</td>
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<td>ω</td>
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<td>24,000±100</td>
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<td>2.7±0.1</td>
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The readings indicate the scatter actually observed in the two tests run for each condition.
### Table 9

Predicted Yield and Ultimate Tensile Strengths

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<tr>
<th>Condition</th>
<th>Condition</th>
<th>Actual</th>
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<th>Present</th>
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<td>$\sigma_f/\sigma_u$</td>
<td>1.05</td>
<td>1.35</td>
<td>1.20</td>
</tr>
<tr>
<td></td>
<td>$\sigma_f/\sigma_{ULYS}$</td>
<td>1.24</td>
<td>1.35</td>
<td>1.20</td>
</tr>
<tr>
<td>16% Cold</td>
<td>$\sigma_f/\sigma_u'$</td>
<td>1.37</td>
<td>1.38</td>
<td>1.27</td>
</tr>
<tr>
<td></td>
<td>$\sigma_f/\sigma_{ULYS}'$</td>
<td>1.30</td>
<td>1.38</td>
<td>1.27</td>
</tr>
</tbody>
</table>
APPENDIX I

Operation of the Texture Goniometer

The texture goniometer (Figure 1) is used to determine the orientation of any family of planes relative to the surface of a flat specimen. The orientations are plotted on a stereographic projection as a pole-figure. The plane of the surface of the specimen and the plane of the page containing the pole-figure are considered to be coincident in these stereographic projections. To determine the relative orientations the sheet material is affixed to the stage of the goniometer and, during an exposure, undergoes three movements:

1. Rotation of the specimen in its own plane.
2. Rotation of the specimen around an axis perpendicular to the axis of the goniometer, and lying in the plane of the specimen.
3. To average the intensities over a large surface of the specimen, a third movement is applied involving a translation of the specimen in its own plane along a line in the axial direction of the goniometer.

A general outline (a detailed explanation follows) of the technique used to produce a pole-figure is now given. The rotations outlined above cause the scattering vector to describe a spiral in the plane of the stereographic projection (Figure 2). Along this spiral the X-ray intensity is measured and recorded (Figure 3). This data is then arranged as a polar diagram (pole-figure), by plotting the appropriate intensities along the spiral of the stereographic projection. Beyond approximately 70° of the vertical angle the specimen becomes too
App. 1-2. Scattering vector due to combined goniometer rotations.
oblique to the X-ray beam for sufficient diffraction to occur.

To cover the range 70° to 90° vertical angle, the sample can be investigated by transmission. Set up this way the specimen undergoes a rotation in its own plane in addition to the sliding movement. The scattering vector now describes a circle rather than a spiral. The azimuth-angle can be set manually in steps of $2^{\frac{1}{2}}\degree$.

The vertical and horizontal angles marked off on the chart of Figure 3 give the co-ordinates at which the corresponding X-ray intensity is plotted (e.g. A; B; C; D; E) on the stereographic spiral of the polar diagram (Figure 2). Before plotting, the intensities are related to the maximum intensity on the chart and are plotted as percentages (see Figure 4). The pole-figure is then produced by drawing contours through the plotted figures.

The relationship between the actual planes, the sheet surface, and the location of the poles of the planes on the pole-figure is shown in Figure 5.

In Figure 5a the relationship between the plane of the X-ray beam and the specimen is shown. During the vertical rotation shown the specimen pivots about the plane of the X-ray beam. This maintains the surface of the specimen in the path of the X-ray beam. The horizontal rotation is about the normal to the specimen surface regardless of the vertical angle. The oscillation occurs regularly regardless of the vertical and horizontal angles.

In Figure 5b it should be noted that the $\phi$ angle denotes the angle between the sheet surface and the diffracting planes in the specimen. Since the sheet surface is the surface of the
1 Plane of X-Ray Beam
2 X-Ray Beam
3 Horizontal Rotation
4 Vertical Rotation
5 Oscillation

stereographic projection the pole of the diffracting planes will be at the centre of the projection for \( \alpha = 0^\circ \) but must be rotated \( \alpha \) degrees from the centre when \( |\alpha| > 0^\circ \). Since the horizontal angle is zero for all three cases the poles move along the transverse diameter of the projection.

Figure 5c shows the effect of varying the horizontal angle when the vertical angle has a value other than \( 0^\circ \). For \( \alpha = 0 \) the pole of the diffracting planes is at the centre of the projection and is unaffected by the horizontal rotation. For the case of \( |\alpha| > 0^\circ \) and constant, the pole will revolve around the circle \( \alpha \) degrees from the centre of the projection. When the horizontal angle \( \beta \) moves to \( 45^\circ \) the pole appears at \( 45^\circ \) above the \( 0^\circ \) position. Similarly for \( 225^\circ \).

The steps followed in producing a pole-figure will be described in detail by using an actual example - Mg + 12.4 atomic % Li which has been cold worked 60% at room temperature. The pole-figure will be for the (0002) planes and CoK\( \alpha \) radiation will be used to limit the depth that is examined in any one test (see the Experimental section for details).

The specimen chosen for examination must be perfectly flat (with parallel top and bottom faces) and have a relatively smooth surface. This may be an as-rolled surface or, if the original surface is rough, it may be ground and electro or chemically polished (to remove grinding effects). The specimen is affixed to the stage of the goniometer with a layer of vacuum grease or plasticene. The specimen is aligned with the rolling direction either in the plane of the X-ray...
beam path or at $90^\circ$ to this plane (the latter orientation allows greater accuracy in alignment). If the rolling direction is set in the plane of the X-ray beam then the horizontal angle scale is set at $0^\circ$; if it is set at $90^\circ$ to the plane, the horizontal angle is set at $270^\circ$ (this is with a negative rotation on the goniometer as shown in Figure 6). The height of the stage is then carefully checked with the probe (Figure 7) by swinging it in an arc across the specimen surface. This also aids in aligning the specimen relative to the X-ray beam if it is initially set at $90^\circ$ to the beam. The height of the specimen is critical if its surface is to be maintained at the centre of the circle of reflection. When the probe is withdrawn, care must be taken to ensure that it does not touch the inner rotating ring of the goniometer to avoid scoring.

The X-ray beam is now turned on and once the voltage and current are adjusted the stage rotation switch is turned on. At the same time the strip chart is turned on. The nature of the stereographic spiral means that the examination of the specimen can be made from $+70^\circ$ to $0^\circ$ or $0^\circ$ to $-70^\circ$. Both scans examine the complete circle.

The procedure followed once the intensity chart is completed is to mark off the angles on the strip chart and then determine the maximum intensity present. Using this as 100% the intensity peaks and valleys are all rated relative to 100%. The intensities corresponding to 10% increments from 0 to 100% are then determined and lines drawn along the chart at these values. The angular readings corresponding to the intersections of these lines and the intensity curve are noted and plotted on the stereographic spiral (see Figures 3 and 4). Once this is completed contours are drawn at 10% intervals on the polar paper to produce the pole-figure.
For the best possible accuracy two corrections may be necessary to the intensities plotted out on the strip chart. These are:

(a) Correction for background -

To determine the amount of background radiation randomly scattered into the counter from the surface of the specimen, the mean value of the intensity obtained (using normal incident beam intensity) with a $2\theta$ angle just above and just below the $2\theta$ angle for the planes being examined is considered to be the background radiation.

(b) Correction for High Vertical Angle -

The goniometer must be carefully aligned relative to the X-ray beam. The goniometer axis must be normal to the plane of the X-ray beam while the vertical axis of the specimen stage must lie in the plane of the X-ray beam for $0^\circ$ vertical angle. To ensure that this is the case a scan is run from $+70$ to $-70$ using a randomly-oriented specimen, such as a powder compact. If alignment is accurate the intensity of the diffracted beam should be constant from $+70$ to $-70^\circ$ with a gradual fall off towards $90^\circ$ at each end.
APPENDIX II

Correlation Between Basal and Pyramidal Pole Figures

The \{\overline{10\overline{1}1}\} first order pyramidal planes were chosen as the most suitable system to use for establishing whether or not any directional texture was present in the materials studied. Since \{1\overline{1}20\} is the active slip direction for the three slip planes that have been observed in Mg (0001) \{10\overline{1}0\} and \{10\overline{1}1\} it will be the direction of interest.

The \{10\overline{1}1\} planes have a high intensity reflection relative to the basal planes (100\% vs 40\%) and their convenient angular relationship (61.9\(^\circ\) from (0001) for pure Mg) mean that reflection techniques can be used rather than the slower transmission method, required if \{10\overline{1}0\} (prism) planes were used.

In hot rolled alloys showing the "ideal" (0001) texture, and no directional texture, the pyramidal planes will have orientations centering around a circle 61.9\(^\circ\) from the centre of the pole figure. For those (0001) planes aligned parallel to the sheet surface the corresponding pyramidal poles will produce a circle at 61.9\(^\circ\) from the centre while basal poles a few degrees away from the surface will produce circles of 61.9\(^\circ\) radius but non-concentric with the centre of the circle. This is shown in Figure A.

When a definite directional texture exists in a material having a near "ideal" texture the pyramidal planes will be grouped in six distinct regions of the pole figure. If the \{1\overline{1}20\} direction tends to align itself with the rolling direction then the pyramidal poles will be grouped as shown in Figure B. This type of curve is shown in Figure 13 surface.

In the cold rolled alloys the majority of the basal planes are not parallel to the sheet surface but occur at approximately
App. II-A. Distribution of pyramidal poles \{10\overline{1}1\} with respect to basal pole.

App. II-B. Distribution of pyramidal poles for material with \(<1\overline{1}20\rangle\) directional texture.
10° from the centre in the positive and negative rolling directions. This means that the pole figure for a randomly oriented direction will be two sets of circles of the type shown in Figure A. Each of these sets will be centred on the 100% peaks of the basal pole figure. An example of this using one circle at 61.9° from each basal peak is given in Figure C. As can be seen rather than producing a pair of circles with equal intensity around their centres it produces a series of peaks and valleys. An example of this is shown in Figure 16b.

When directional texture is present in a (0001) split-texture pole figure each of the 1011 circles should show six distinct peaks as described for the "ideal" type texture. This is shown in Figure D. No examples of this were observed in the present work.
App. II-C. Typical pole figure actually obtained.

App. II-D. Expected pole figure for (0001) split texture with directional texture.
APPENDIX III

Definition for Terms Used to Describe Mechanical Properties

The following definitions are given for the terms used in the description of the mechanical properties of Mg and Mg-Li alloys:

\[ \sigma_{YS} = \text{the 0.2\% offset yield strength} \]
\[ \sigma_{YS} = \frac{\text{Load at 0.2\% offset}}{\text{Original Area}} (1.002) \]
\[ \sigma' = \text{the ultimate tensile strength} \]
\[ \sigma' = \frac{\text{Load at Maximum of Load-Elongation Curve}}{\text{Original Area}} (1 + \varepsilon_u) \]
\[ \varepsilon_u = \text{Uniform strain} \]
\[ \varepsilon_u = \frac{\text{Elongation to the beginning of necking}}{\text{gauge length}} \]