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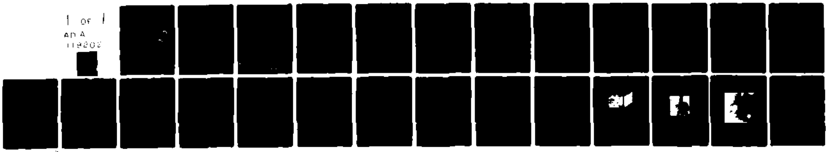

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"Stress Corrosion Cracking of Wrought and P/M High Strength Aluminum Alloys"

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Principal Investigators: A. W. Thompson, I. M. Bernstein





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**20. ABSTRACT (Continue on reverse side if necessary and identify b. block number)**
This report presents results for the first year of a three-year program. The objective of this program is to understand the stress corrosion cracking (SCC) behavior of high-strength ingot and powder metallurgy aluminum alloys, and in particular to understand the role of hydrogen in such cracking. The approach taken was to study microstructural effects on both hydrogen embrittlement and SCC, and to establish, insofar as possible, microstructural and fractographic correlations with cracking behavior, and detailed understanding

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SECURITY CLASSIFICATION OF THIS PAGE (When Data Entered) of the mechanical behavior of hydrogen-charged material. First-year results are summarized as follows:

7075 Al with the T6-RR temper appears to be no better than the conventional T6 temper in constant load, 3.5% NaCl solution SCC tests. SET testing, where large hydrogen concentrations are introduced, seems to confirm this.

In both peak aged and overaged conditions, the copper-containing 7075 alloy was unaffected by hydrogen even under the severe hydrogen entry conditions of the SET test.

In general, embrittlement results for 2124 seem to conform to those established for 7075: the underaged microstructures are more susceptible to hydrogen embrittlement.

For 7075, SCC resistance increases as aging proceeds from UT to T6 T73. This has been demonstrated in both constant load and SET testing.

For short transverse specimens of 7075 Al, Mode I specimens failed sooner than Mode III for all three tempers, indicating that hydrogen plays a major role in the SCC process.

For longitudinal specimens of 7075, Mode I specimens failed sooner than Mode III in T6 temper, but vice versa for both UT and T73 tempers, indicating a strong grain orientation effect.

Embrittlement (RA loss) of X-7090 by cathodic precracking suggests a strong hydrogen affect in this alloy. Mode I/Mode III SET tests tend to confirm this.

## 1.0 Abstract of Results

This report presents results for the first year of a three-year program. The objective of this program is to understand the stress corrosion cracking (SCC) behavior of high-strength ingot and powder metallurgy aluminum alloys, and in particular to understand the role of hydrogen in such cracking. The approach taken was to study microstructural effects on both hydrogen embrittlement and SCC, and to establish, insofar as possible, microstructural and fractographic correlations with cracking behavior, and detailed understanding of the mechanical behavior of hydrogen-charged material.



## 2.0 Technical Results

The intent of the overall program is to identify the role played by hydrogen in stress corrosion cracking by establishing microstructural and fractographic correlations with cracking, and by mechanical tests under cathodic charging conditions. Specific emphasis has been placed on the role of such metallurgical variables as precipitate type, size, and distribution, in order to develop them as control variables for the production of wrought and powder high-strength aluminum alloys more resistant to hydrogen embrittlement (HE) and/or stress corrosion cracking (SCC). First-year results are summarized in the following sections.

### 2.1 Stress Corrosion Cracking of 7075-Mode I/Mode III Testing

Having obtained an understanding of the internal hydrogen embrittlement behavior of 7075 under tensile loading during the previous 4-year program (1), the emphasis shifted to stress corrosion cracking (SCC) and stress state tests on the same microstructures. The primary procedure used was to vary loading mode, a concept which can furnish a means of identifying the proportion of SCC susceptibility which arises from a hydrogen contribution (2). Because of the difference in stress state at the crack tip for the two modes, hydrogen's role in the fracture process can be more easily identified. Specifically, hydrogen will more easily accumulate in the tensile stress field of a Mode I crack thereby exacerbating any deleterious effect it may have. In pure Mode III additional hydrogen should not accumulate and no enhancement should be observed. A detailed analysis of such results of 7075 in three temper conditions (peak aged, underaged, and overaged) was provided previously (1).

The loading mode study was also extended to 7075 aluminum with a relatively new temper which has been reported (3) to reduce the SCC susceptibility of 7075-T6 aluminum, while maintaining its mechanical properties. This temper, called retrogression and re-aging (RR), begins with a conventional T6 temper, adds a retrogression step at a temperature between 205°C and 260°C for a short time, followed by a water quench, and re-aging at the original aging temperature of 120°C. The RR treatment is based on the hypothesis (4) that localized populations of dislocations associated with undissolved $MgZn_2$ particles are the primary sites for SCC in 7075 Al. Testing reported by other researchers (3,5) involved U-bends and constant elongation tensile specimens in either alternate immersion testing or total immersion in boiling 6% NaCl solutions. These reports stated that the T6-RR temper has two to three times longer times to failure than the conventional T6 temper, when loaded to about 75% of their respective yield stress. We decided to test this prediction in a 3.5% NaCl solution, using a constant load apparatus.

Time-to-failure results for notched longitudinal specimens having the T6-RR temper are shown in Figure 1, with the conventional T6 temper results included for comparison. In Mode I, tne T6-RR temper appeared to be slightly better than the T6 temper. However, the Mode III results showed mixed results showed mixed results: at low applied stress intensity, T6 is better, and at higher applied stress intensity, T6-RR is better. It would appear that the RR treatment does not significantly affect SCC resistance, particularly in notched specimens where the need for initiation has been reduced. These results have been detailed and fully evaluated in a recent publication (6).

2.2 Straining Electrode Tests in 7075

As described in previous technical reports (1) and publications (7), the straining electrode test (SET) technique, which combines tensile loading with continuous cathodic charging, accelerates the effect of hydrogen on environmental cracking and can serve as an important diagnostic tool in assessing how serious the hydrogen effect can be in a given alloy or temper.

The SET procedure was conducted with notched specimens in Mode I loading in pH 1 HCl solutions at room temperature, with a cathodic overpotential of 150 mV; the crosshead rate was about $10^{-4}$cm/s. The SET results for longitudinally oriented samples showed fracture stress intensity values similar to those for specimens tested in air for both the T6 and T73 tempers, as shown in Table I. The fracture stress intensity value for the UT temper, however, dropped about 22% in the SET, suggesting more severe embrittlement in the UT temper than in the others, following the trends found previously in both precharged and NaCl tests (1). Longitudinally oriented T6-RR specimens also showed a pronounced drop in fracture stress intensity that was not present for the T6 temper. This difference has not yet been explained, but its presence again does not support the contention that this temper improves environmental resistance.

The short transverse oriented specimen showed nearly the same large differences between air and SET toughness values for all three conventional tempers, Table I, demonstrating again the importance of grain shape (7). It is quite clear that overaging cannot be a guaranteed inhibitor of environmental embrittlement without proper control of other variables.

In order to establish the maximum effect of hydrogen charging on the T6 and T73 tempers, which showed little embrittlement in the longitudinal orientation in our standard SET tests, the strain rate was lowered in an attempt to sweep in more hydrogen via mobile dislocations (7). To accomplish this the crosshead rate was reduced to about $10^{-5}$cm/s. The resulting toughness values are shown in Table I. The value for the T73 temper is essentially unchanged, while the T6 temper showed a 28% loss compared with the air test. Thus, the SET technique again shows that both the UT and T6 tempers are more susceptible to environmental cracking than the T73 temper. Also, the deleterious effect of a decreasing strain rate has again been demonstrated (8).

Table I

Fracture Stress Intensity Values for Notched Specimens

| | Longitudinal Orientation | | | Short Transverse Orientation | | |
|---|---|---|---|---|---|---|
| | Air | SET | % Loss | Air | SET | % Loss |
| UT | 25.1 MPa$\sqrt{m}$ | 19.7 MPa$\sqrt{m}$ | 22 | 22.0 | 14.7 | 33 |
| T6 | 25.1 | 25.1 | 0 | 22.5 | 15.4 | 32 |
| | | 18.0 (slow) | 28 | | | |
| T73 | 22.4 | 21.7 | 3 | 20.3 | 14.5 | 28 |
| | | 22.4 (slow) | 0 | | | |
| T6-RR | 30.0 | 26.0 | 13 | -- | -- | |

## 2.3 Hydrogen Embrittlement of 2124 Al

Work has also been completed on an alloy of the 2000-series (based on Al-4% Cu) (9). The 2000-series alloys share an important behavioral characteristic with their high-strength 7000-series counterparts in that both are susceptible to SCC (1,10). Consequently, as an extension of the work on 7075, the susceptibility to hydrogen of a 2000-series alloy, 2124 (Cu-4.8%, Mg-1.46%, Fe-0.74%, Mn-0.5%, Si-0.099%, Ti-0.19%, Ni-0.01%) was examined.

Details of the work were provided previously (1), and a recent publication (9) provided a detailed analysis of results. In general, it was demonstrated that 2124 could be embrittled by cathodically-charged hydrogen, provided that hydrogen entry conditions were severe. In general, the results conformed to the framework established for the 7075 alloy: the underaged microstructures were the ones susceptible to hydrogen embrittlement, and this can in general terms be understood as a microstructure and slip behavior effect.

### 2.4 Influence of Copper on the Hydrogen Embrittlement of 7xxx Aluminum Alloys

In the course of the past year, a study was begun and largely completed on the influence of copper additions on the hydrogen embrittlement of 7000 type high-strength Al-Zn-Mg alloys. It is well known that the addition of copper to high strength Al-Zn-Mg alloys simultaneously improves their yield strength and SCC resistance (11). In fact, an overaging heat treatment may be effectively used to greatly enhance stress corrosion resistance without unacceptable strength penalties in copper-containing alloys.

The compositions of the two alloys examined (essentially 7050 and Cu-free 7050), which were obtained from Alcoa Research Laboratories, are given in Table II. Both alloys, obtained as extruded plate, had a predominantly "pancake" grain shape; Zr was the element responsible for the inhibition of grain growth through the presence of $Al_3Zr$ precipitates. However, some degree of recrystallization had occurred so the material essentially had a dual grain size.

As was done previously, loss of ductility in the tension test was used as the measure of hydrogen embrittlement. Specimens were tested in the uncharged and cathodically-charged conditions and straining-electrode tests (SET) were also performed. Reduction-in-area (RA) results obtained for both alloys in various heat treatment conditions (Table III) are shown in Figure 2. Also included in this figure are results obtained with underaged commercial 7075 and peak aged (T8) 2124.

The most significant result obtained was that in both the peak and overaged conditions, the copper-containing alloy (7050) was unaffected by hydrogen even under the severe hydrogen entry conditions generated by the straining electrode test. Analysis of these results is continuing in order to elucidate the role of copper in promoting hydrogen embrittlement resistance. Such studies include analysis of local copper partitioning at grain boundaries and precipitates using high resolution scanning transmission electron microscopy (STEM) techniques.

Table II

Composition of Al-Zn-Mg and 7050 Alloys*

| | Zn | Mg | Cu | Fe | Ti | Si | Zr | Al |
|---|---|---|---|---|---|---|---|---|
| Al-Zn-Mg** | 6.42 | 2.06 | 0.01 | 0.04 | 0.02 | 0.05 | 0.11 | Bal |
| 7050 | 5.92 | 2.10 | 2.10 | 0.07 | 0.02 | 0.06 | 0.12 | Bal |

*Both alloys were free of Mn, Cr, Ni and Be.

**Intended as essentially Cu-free 7050.

Table III

*Heat Treatment Conditions*

Alloy: Al-Zn-Mg.

| Condition | Treatment | Strength (MPa) |
|---|---|---|
| Underage | 10hrs/100$^{o}$C | 400 |
| Peak-age I | 24hrs/120$^{o}$C | 480 |
| Peak-age II | 48hrs/RT + 24hrs/120$^{o}$C | 480 |

Alloy: 7050

| Condition | Treatment | Strength (MPa) |
|---|---|---|
| Underage I | 10hrs/100$^{o}$C | 460 |
| Underage II | 24hrs/120$^{o}$C | 510 |
| Peak Age | 24hrs/120$^{o}$C + 10hrs/150$^{o}$C | 550 |
| Overage | 24hrs/120$^{o}$C + 32hrs/150$^{o}$C | 530 |

## 2.5 Environmental Cracking of P/M X-7090

The X-7090 alloy used in the present study was provided to us in the peak-aged condition by ALCOA. The fabrication procedure is listed in Table IV. Table V shows the nominal composition of X-7090 with the I/M 7075 included for comparison. Although the composition of these two alloys are somewhat similar, the aging sequence is likely different for X-7090 than for 7075 because of the increased Zn concentration.

Figure 3 shows the grain structure at the center of the as-received extrusion. Transmission electron microscopy (TEM) has shown precipitate-free zones adjacent to grain boundaries which themselves are decorated with large $\eta$ precipitates, as in Figure 4. Spherical $Co_2Al_9$ or $(CoFe)_2Al_9$ precipitates (12) are found at grain boundaries and in the matrix. Rectangular secondary intermetallic compounds and oxide dispersions lie within the grains. Figure 5 shows some of these features.

Table IV

Fabrication Procedure for Al X-7090

1. Air atomize powder.
2. Sieve to -325 mesh (less than 44 μm).
3. Isostatic press to 75% TD.
4. Can in 3003 Al, evaporate.
5. Soak at 521°C, 1 hr.
6. Hot press to 100% TD, 521°C.
7. Extrude at 370°C, reduction ratio of 24:1.
8. Solution heat treat 488°C.
9. Cold water quench.
10. Stretch 1½% (to relieve residual quench stresses).
11. Room temp. age 4 days.
12. Artificial age 121°C 24 hr.

Table V

Alloy Composition

| | | Zn | Mg | Cu | Co | Fe | Si | Mn | Ti | O | Cr |
|---|---|---|---|---|---|---|---|---|---|---|---|
| P/M | X-7090 | 7.0-8.4 | 2.0-3.0 | 0.6-1.3 | 1.2-1.7 | 0.15 | 0.12 | ---- | ---- | 0.20-0.50 | --------- |
| I/M | 7075 | 5.1-6.1 | 2.1-2.9 | 1.2-2.0 | ------- | 0.50 | 0.40 | 0.30 | 0.20 | --------- | 0.18-0.28 |

A slip characterization study has been conducted to complement the matrix precipitate study to help demonstrate whether a slip mode contribution to SCC can be obtained and then altered in these alloys. This study included examination of polished surfaces of compression specimens as well as TEM foils at various depths from the surface. The lack of success to date in characterizing slip in the peak aged material is considered to result from the incompletely recrystallized microstructure as well as the fine grain size (about 5 μm diameter). Another technique, TEM examination of replicas, will be employed in an attempt to more fully characterize slip in this alloy.

In order to assess the effect of aging treatment on yield strength, cubes of X-7090 Al, measuring about 1 cm on a side, were heat treated and compressed. The results are shown in Figure 6, with 7075 Al results shown for comparison. The large increase in strength from room temperature aging for four days suggests the presence of a substantial amount of precipitation. However, a TEM investigation of room temperature aged X-7090 has failed thus far to identify these precipitates, likely owing primarily to their small size. Also of interest in Fig. 6 is the high degree of strength attained by aging at 100°C for 24 hours. Aging at high temperatures tends to decrease the yield strength, even more quickly than for 7075.

Smooth, round, tensile specimens of X-7090 were precharged with hydrogen for 10 h under an applied constant potential of -1500 mV vs. standard calomel electrode in a hydrochloric acid solution (pH 1). These results are shown in Table VI, along with similar results from 7075 similarly tested (1,13). The higher RA loss in X-7090 indicates the hydrogen effect may be even greater for X-7090 than for 7075.

Notched round specimens of X-7090 were cathodically charged under a constant potential of -1500 mV in hydrochloric acid solution (pH 1) while being strained to failure in Mode I or Mode III loading. This comprised a Mode I/Mode III

TABLE VI

TENSILE TESTS: EFFECT OF HYDROGEN CHARGING

| Age | P/M X-7090 RA Loss | I/M 7075 RA Loss |
|---|---|---|
| Room - Temp. (2 days) | 45% | -- |
| Peak Age (as-received) | 57 | 31 |
| $160^{o}C$ (24 hr) | 40 | 16 |

$$\text{RA Loss} = \frac{RA - RA_{hydrogen}}{RA}$$

Charged 10 hr. in pH 1 HCl, - 1500 mV SEC

study to introduce hydrogen under the severe entry conditions of the SET test. These results are shown in Table VII with stress intensities for SET testing normalized to their air test values.

Since normalized Mode I values are lower than the normalized Mode III values, these results also suggest that hydrogen plays a major role in the SCC of X-7090 in the peak aged condition, again in contrast to 7075.

TABLE VII

Mode I/Mode III SET Results

| | Mode I | Mode III |
|---|---|---|
| | $K_{I,SET}/K_{I,AIR}$ | $K_{III,SET}/K_{III,AIR}$ |
| Peak Aged | 0.74 | 1.13 |

Several additional types of tests will be conducted to assess the role of hydrogen in the SCC of this alloy. These include loading mode, potential dependence on crack growth rate, and fractography.

The loading mode tests will use peak-aged, underaged, and overaged specimens, as in ref. 13 and 14, to serve as an initial indicator of a hydrogen contribution. If a change in slip mode can be demonstrated by further analysis for different tempers, this is expected to change the time-to-failure curve differences between Mode I and Mode III loadings for different tempers.

Double cantilever beam specimens are being prepared to be tested in HCl solutions at various potentials to assess the effect of bulk potential on V-K curves. While it is unlikely that this study will produce a valid hydrogen mechanism to account for crack propagation rates from the V-K tests, the crack propagation rates from these tests will be compared with those from a series of slow strain rate tests conducted over a range of potentials. If calculated anodic dissolution rates are comparable with crack growth rates from these tests, then the fracture mode may indicate a dissolution mechanism. If the fracture mode is intergranular, grain boundary analysis by scanning transmission electron microscopy (STEM) will be conducted to check for solute depletion of for example, Cu, to help explain the IGSCC in terms of a localized dissolution.

## 2.6 Further Details on Ongoing Research

2.6.1 The proposed work on the wrought alloy 7075 will have lower priority than that on alloy X-7090. Some further work will be undertaken in order to determine whether some sort of recovery of ductility exists in hydrogen charged and embrittled 7075 alloy during slow strain rate tensile testing. This work may help determine whether the recovery in ductility in hydrogen embrittled 7075 is due to trapping of hydrogen at internal sites or to the loss of hydrogen to the surrounding environment. If hydrogen is escaping from the specimen during the slow strain rate test, the recovery of ductility should be related to the hydrogen pressure in the surrounding environment of the specimen. Thus, the amount of recovery (the increase in ductility, due to hydrogen loss) would be inversely related to hydrogen pressure. Recovery of ductility will be measured in terms of loss in reduction of area (RA-loss) compared to reference tests carried out with uncharged specimens in eithe a dry argon environment or in vacuum.

Commercial IM 7075 sheet material will be used for this work. Tensile specimens will be machined parallel to the rolling direction, in order to compare the results with earlier work done (1) on this alloy. All specimens will be given a full heat treatment after machining to achieve the T6 condition (peak age). Heat treated specimens will be charged in 0.1M HCl (pH 1) for 10h at -1500 mV (SCE) and stored in liquid nitrogen. Tensile tests will be carried out with charged and uncharged specimens in a vacuum chamber under different hydrogen pressures ($1 \times 10^{-6}$ to 2 atm). The hydrogen gas used will be dried carefully to avoid any embrittlement due to moisture, which can occur at slow strain rates and thus influence markedly the recovery behavior of the material. Scanning electron microscopy will be used to examine fracture surfaces of charged and uncharged specimens.

2.6.2 In order to study fundamentally the hydrogen embrittlement response and related mechanisms in the PM alloy X-7090, it will be necessary to increase the grain size of this alloy to achieve some sort of "model alloy", with an equiaxed grain structure. Therefore, a recrystallization study will be undertaken with the PM alloy X-7090. This work will give us information about the dependence of the grain size on recrystallization temperature and degree of deformation. Additional information on the stability of the microstructure at high temperature exposure will be worked out by a transmission electron microscope examination.

Grain size studies will begin by examining strain growth behavior at high temperatures, followed by recrystallization experiments. Samples will be deformed to different degrees and heat treated near the solution temperature for various times. Time dependence and also deformation dependence of the recrystallized grain size will be measured in order to define data for further work, where specimens with a certain grain size would be required.

In connection with this recrystallization study, it will be possible to examine the microstructural development of PM X-7090 after exposure to high temperatures. Therefore, transmission electron microscopy investigations will be made to describe the change in structure and to examine all phases present.

2.6.3 Additional tensile specimens of X-7090 will be cathodically charged with hydrogen before pulling to failure. These fracture surfaces will be compared with those from the SCC tests. Strain rates will be altered to determine whether the fracture can be forced from being discontinuous to continuous, or vice versa. The Mode I/Mode III SET tests will be extended to other tempers. In addition, some TEM will be conducted on fracture surfaces to check for presence of hydrides (15). While it is possible that our results in X-7090 will show SCC response similar to that of 7075, the somewhat different microstructure of X-7090 will provide a suitable test candidate for evaluating the microstructure-environment interaction. Perhaps the key contribution of this work will be to extend current understanding of SCC, by separating hydrogen embrittlement from anodic dissolution and by demonstrating microstructural control of SCC.

## 2.7 Conclusions

- 7075 Al with the T6-RR temper appears to be no better than the conventional T6 temper in constant load, 3.5% NaCl solution SCC tests. SET testing, where large hydrogen concentrations are introduced, seems to confirm this.
- In both peak aged and overaged conditions, the copper-containing 7050 alloy was unaffected by hydrogen even under the severe hydrogen

entry conditions of the SET test.

- In general, embrittlement results for 2124 seem to conform to those established for 7075: the underaged microstructures are more susceptible to hydrogen embrittlement.
- For 7075, SCC resistance increases as aging proceeds from UT to T6 T73. This has been demonstrated in both constant load and SET testing.
- For short transverse specimens of 7075 Al, Mode I specimens failed sooner than Mode III for all three tempers, indicating that hydrogen plays a major role in the SCC process.
- For longitudinal specimens of 7075, Mode I specimens failed sooner than Mode III in T6 temper, but vice versa for both UT and T73 tempers, indicating a strong grain orientation effect.
- Embrittlement (RA loss) of X-7090 by cathodic precracking suggests a strong hydrogen affect in this alloy. Mode I/Mode III SET tests tend to confirm this.

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## 5.0 Interactions

### 5.1 Presentations

1. A. W. Thompson, "The Role of Microstructure in Environmentally-assisted Fracture", Materials Durability Center Seminar, University of Delaware, Newark, DE, 24 April 1981.

2. I. M. Bernstein and A. W. Thompson,"Transient Effects During Hydrogen Transport by Dislocations", NATO Advanced Research Institute, Atomistics of Fracture, Calcatoggio, Corsica, 28 May 1981.

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5. A. W. Thompson and I. M. Bernstein, "Quantitative Metallography and Fractography in Hydrogen Embrittlement Studies", ibid., 13 October 1981.

6. A. W. Thompson and I. M. Bernstein, "Micromechanics of Environmentally-Assisted Crack Growth", Symposium on Micro and Macro Mechanics of Crack Growth, Fall Meeting, TMS-AIME, Louisville, KY, 13 October 1981.

7. A. W. Thompson and I. M. Bernstein, "Stress Corrosion Cracking of P/M Aluminum Alloys", AFOSR Working Group Meeting on P/M Aluminum Alloys, Louisville, KY, 15 October 1981.

### 5.2 Technical Contacts with Other Investigators

The principal investigators listed above have had extensive contact with other AFOSR investigators in related fields, as briefly summarized below.

Prof. W. W. Gerberich, University of Minnesota: Discussions on hydrogen accumulation at precipitates and grain boundaries in aluminum alloys. General discussions about aluminum alloy fracture modes and stress corrosion cracking phenomena.

Prof. J. L. Swedlow (solid mechanics) Carnegie-Mellon University: Discussions of microstructural aspects of deformation and fracture in two-phase alloys.

Prof. J. C. Williams, Carnegie-Mellon University: Discussions on strengthening mechanisms in aluminum alloys.

We also had the opportunity to meet with the entire AFOSR-sponsored P/M aluminum working group at the Louisville fall meeting of TMS. Also in attendance were other Air Force representatives such as from Wright Patterson. This provided a valuable opportunity for important discussions.

## 5.3 Personnel

A. W. Thompson - Professor and Co-Principal Investigator (20% AY, 50% summer).

I. M. Bernstein - Professor and Co-Principal Investigator (5% AY, 30% summer).

D. Hardwick - Post-doctoral Associate (terminated 31 July 1981).

R. E. Swanson - Graduate Student.

M. Mueller - Post-doctoral Associate

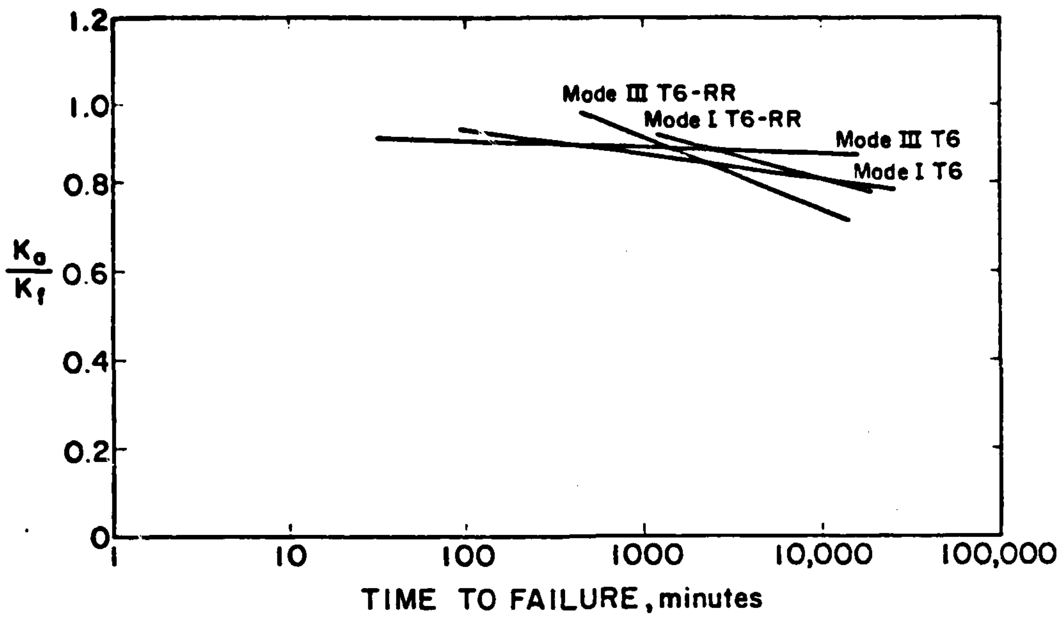



FIGURE 1. Dependence of time to failure on stress intensity ratio (applied K divided by K for failure with rising load) for notched longitudinal specimens of 7075 with either T6 or T6-RR tempers, tested in aqueous 3.5% NaCl solution at $50^{0}$ C. Tests conducted in both Mode I (tension) and Mode III (torsion).

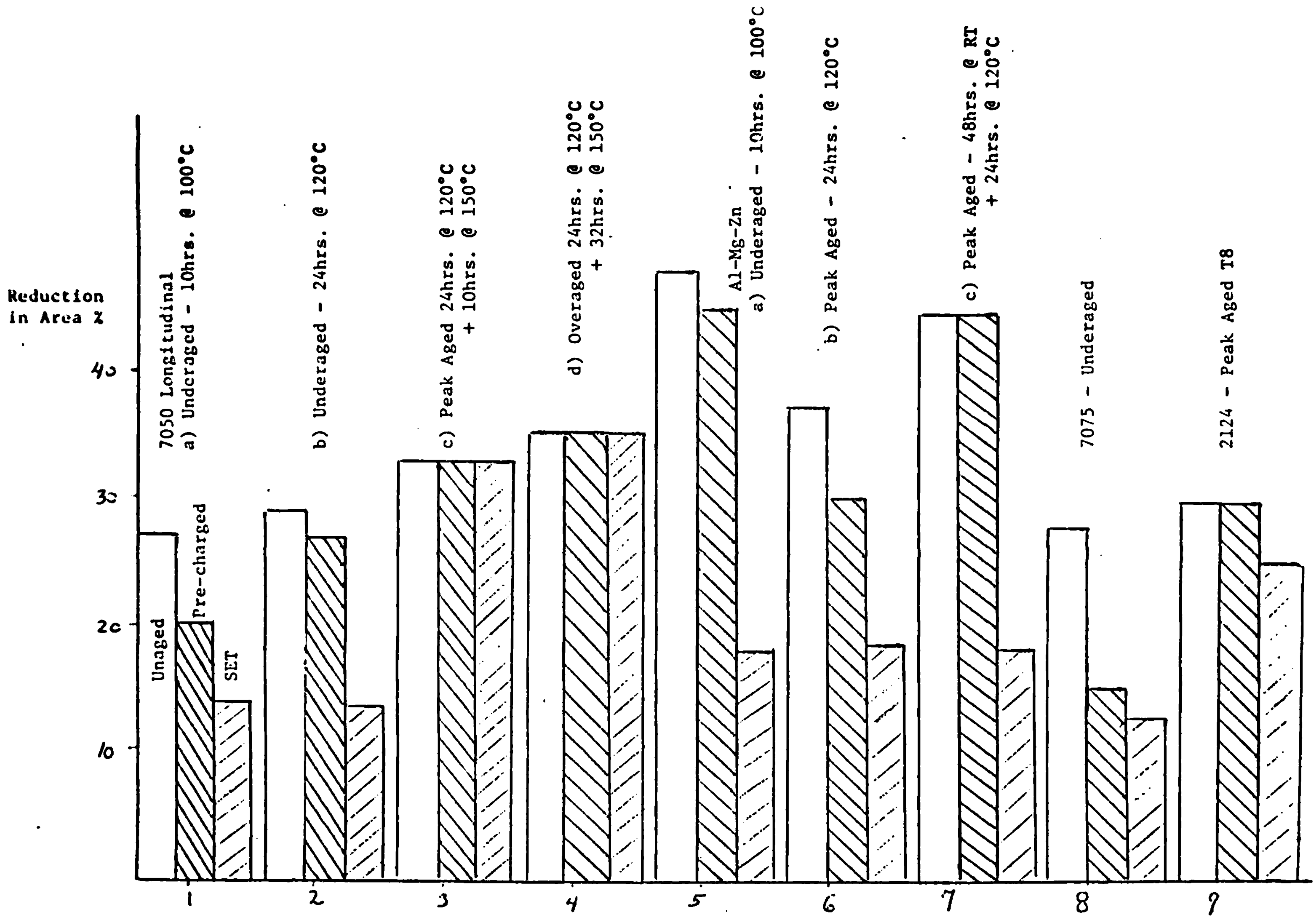



Figure 2: Reductions in Area for Aluminum Alloys of Various Tempers.

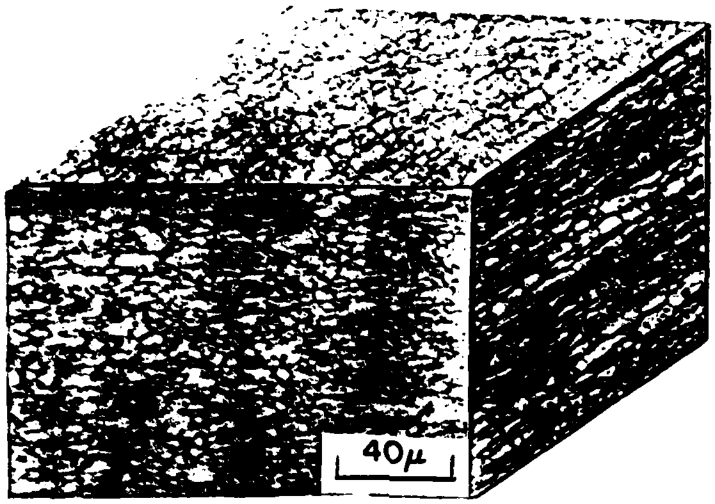
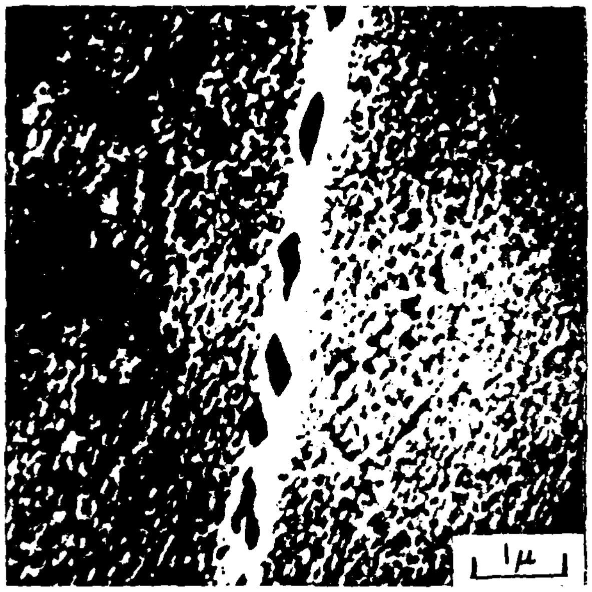



FIGURE 3

GRAIN STRUCTURE AT CENTER OF EXTRUSION




7090 P/M Al PFZ WITH ETA AT GRAIN BOUNDARY

FIGURE 4


7090 P/M Al AGED 24 HR AT 100C

RECTANGULAR PARTICLE RICH IN Fe, Si

ROUND PARTICLE $Co_2Al_9$

FIGURE 5

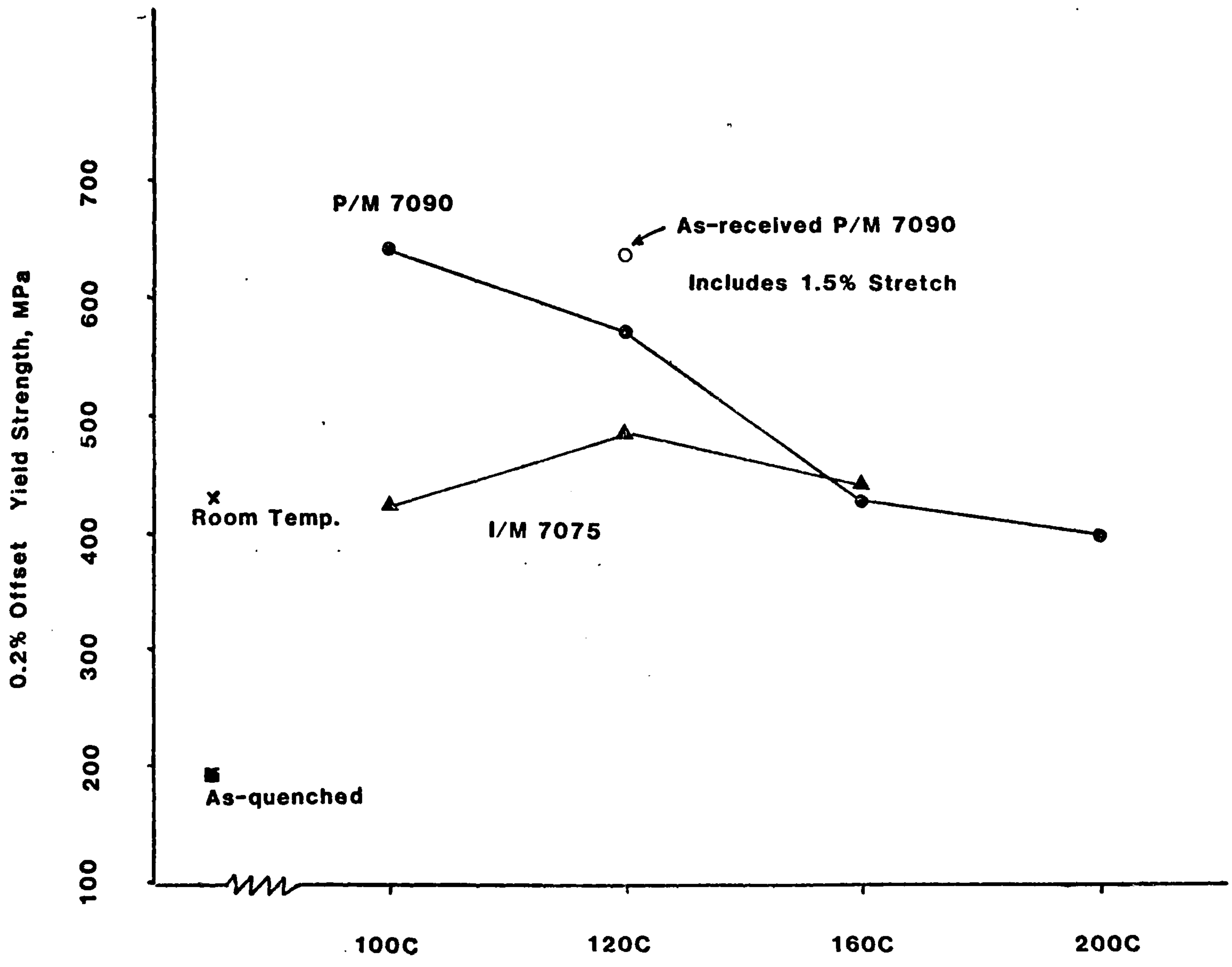



Figure 6. **Age Treatment**

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