Thermomechanical Treatments on High Strength Al-Zn-Mg(-Cu) Alloys

E. DI RUSSO, M. CONSERVA, F. GATTO, AND H. MARKUS

An investigation was carried out to determine the metallurgical properties of Al-Zn-Mg and Al-Zn-Mg-Cu alloy products processed according to newly developed Final Thermomechanical Treatments (FTMT) of T-AHA type. The results show that these cycles can be utilized to produce wrought products of high purity Al-Zn-Mg(-Cu) alloys characterized by equivalent toughness and ductility and much higher strength than conventionally processed commercial purity materials. Based on transmission electron microscopy studies, it was found that such improved behavior of FTMT material is attributable to the superposition of hardening effects, from aging precipitation and from dislocations. Preliminary stress-corrosion and fatigue tests indicate that these properties are not substantially influenced by T-AHA thermomechanical process. Further work is needed in this area, in order to better understand the directions to follow for developing better alloys.

RECENTLY, increasing recognition has been given to thermomechanical treatments (TMT) as important techniques of improving the properties of metallic materials. The use of such treatments has been limited primarily to iron base alloys and therefore it has been generally accepted that the term TMT refers to those treatments which involve deformation prior to or during phase transformations.[1] In addition to this definition all of the metallurgical processes based on the combination of plastic deformation and heating, but not involving allotropic changes, have been reported as mechanical-thermal treatments (MTM).[2] However, other types of alloy/treatments exist which cannot be classified under the strict definition of TMT but for which the term MTM is too general. Such a situation is represented by processes comprising plastic deformation and precipitation phenomena in aluminum alloys. The structural changes involved in these treatments cannot be considered as simple rearrangements of dislocations because the interaction between the lattice defects and the decomposition of the solid solution is characterized by complex synergistic effects. For this reason it seems reasonable to use the term TMT for those treatments on aluminum alloys that involve combination of plastic deformation and precipitation processes to give new strength-structure relations, even though aluminum alloys do not exhibit polymorphic transformations and as such do not strictly fall within the definition of TMT as reported in Ref. 1.

In addition, in aluminum alloys, different types of precipitation phenomena occur. The principal ones are: the high temperature precipitation of ancillary elements, which is usually produced during the thermal processing of the cast material, and the low temperature precipitation of solute rich particles, which occurs during the age-hardening stage. In order to give better clarity to the work, for thermomechanical cycles of aluminum alloys involving precipitation of ancillary elements we have proposed the use of the term ITMT (Intermediate Thermomechanical Treatments), whereas when low temperature age-hardening phenomena are involved, the term FTMT (Final Thermomechanical Treatments) is suggested.

In the present work, the application of FTMT to high strength Al-Zn-Mg(-Cu) alloys has been investigated; our goal was to establish the conditions in which the dislocations introduced by plastic deformation interact most favorably with the age-hardening processes to produce materials that have improved properties compared to conventionally aged materials.

1) APPROACH

An overall view of the tensile properties of the most widely used commercial Al-Zn-Mg(-Cu) alloys is shown in Fig. 1. It can be seen from this figure that with the standard processed alloys, yield strength (YS) levels of 90,000 psi can be approached with elongation values of about 10 pct. Recently, considerable effort has been directed towards improving the strength of wrought Al-Zn-Mg(-Cu) alloys and some investigators produced these alloys with YS values that were higher than 100,000 psi, Fig. 1. Various combinations of strength and elongation have been achieved by Di Russo[3] by controlling the solidification conditions, chemical composition and the homogenization, solution treatment and aging cycles. Flemings and coworkers[4] obtained similar results on alloys that were solidified under controlled conditions, completely homogenized and then cold worked (H) after normal solution (T) and aging (A) treatments. The use of different combinations of aging and work hardening was tested by Mercier *et al.*[5] using a process based on plastic deformation after complete T6 heat treatment, followed by a partial recovery treatment. Studies were also made by Pavlov *et al.*[6] in which the solution treatment was followed by warm deformation (H_C) and then by artificial aging (A).

From Fig. 1 it can be seen that the increase in strength is always accompanied by a decrease in elongation which, at the highest YS level, drops to a very low value. This is more apparent for the treatments involving combinations of deformation and aging. Neglecting the T-AH treatments, in which no signifi-

E. DI RUSSO and M. CONSERVA are Research Metallurgists and F. GATTO is Director, the Istituto Sperimentale dei Metalli Leggeri, Novara, Italy. H. MARKUS is Chief, Materials Engineering Division, Pitman Dunn Laboratory, Frankford Arsenal, Philadelphia, Pa. 19137.

Manuscript submitted June 26, 1972.

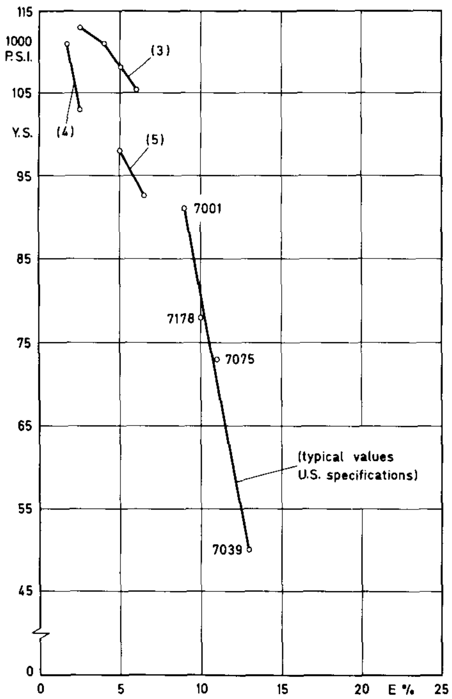


Fig. 1—Yield strength values as a function of elongation shown by commercial and experimental alloys of Al-Zn-Mg(-Cu) system. The numbers in parentheses indicate the reference numbers.

cant relation exists between the two factors A and H, it is well known that the sequence T-HA in Al-Zn-Mg(-Cu) alloys, which involves plastic deformation after solution treatment, leads to a reduction in the response to subsequent artificial aging.[7,8]

In fact, the two parameters H and A are competitive so that properties of materials in the T-HA state are not substantially different from those of the TA state.[9-11] However, developments in the knowledge of the aging mechanisms and of the microstructural features of the Al-Zn-Mg(-Cu) alloys,[12-18] have suggested that it should be possible by means of more sophisticated TMT to modify plastic deformation and precipitation phenomena to get improvements in strength without drastic loss of ductility. To this end it was felt that the inhibiting effect that dislocations exert on the aging reactions could be substantially reduced by an initial aging step at a low temperature which would be given prior to the plastic deformation and the final aging step (T-AHA cycles).

This type of TMT has not been extensively studied but there are some data on Al-Cu and Al-Mg-Si alloys.[8,19,20] Our preliminary investigations in 1967 on Al-Zn-Mg alloys[21-23] showed that TMT involving solution, quenching, aging, working, aging (T-AHA), markedly increased the ultimate tensile strength (UTS) and YS values of these alloys compared to the conventional tempers, while still maintaining adequate ductility.

More recently, extensive research at our laboratory on Al-Zn-Mg(-Cu) alloys confirmed that these new TMT (T-AHA)* produce alloys that have structures with improved properties compared to conventionally heat treated materials.[24-31] In addition T-AHA cycles were shown to have a beneficial effect on other heat treatable aluminum alloys.[32,33]

*It. Pat. No. 886,185. U.S. Filed 14.4.70 Ser. No. 28,514.

This paper presents a summary of the data obtained on the effect of TMT on the structure-property relationships in Al-Zn-Mg(-Cu) alloys. In more detail, selected results showing the tensile properties, notch and fracture toughness, fatigue and stress-corrosion behavior of different alloy/treatments are presented and discussed. Other papers will be published in the near future dealing with the more fundamental aspects of TMT.

2) EXPERIMENTAL PROCEDURE

a) Material Selection

The basic guidelines used for selecting the materials to study were: i) to have Al-Zn-Mg(-Cu) alloys with a wide range of solute contents. Thus, the alloys studied were 7005, 7039, and 7075. The composition of the alloys investigated and the types of wrought products used are shown in Table I; ii) to have industrially produced and laboratory produced alloys. The industrially produced alloys were all of commercial purity and were prepared as either slabs or ingots (alloy Nos. 1, 2, 4, 5). The laboratory produced alloys were prepared as both high purity (alloys Nos. 3, 7) and commercial purity (alloy No. 6) in the form of semicontinuous dc cast 110 mm diam ingots using controlled solidification conditions.[34,35] The typical microstructures of the laboratory produced ingots in both the as-cast state and after homogenizing at 450°C for 8 h + 480°C for 15 h are shown in Fig. 2. The dendrite arm spacing (DAS) of the alloys range from 20 to 30 μ. Due to the small DAS, the homogenization treatment used was sufficient to dissolve all of the eutectic network present in the high purity alloy ingots. In the alloy of commercial purity (alloy No. 6 : c. p. 7075) some of the secondary compounds containing the impurity elements, Fe and Si, combined with Al and Cu, or Al, Cu, and Mg remain undissolved. In the industrially produced commercial purity alloys, the solidification rates are lower and hence many coarse undissolved phases, containing iron and silicon, remain after the homogenization treatment. These microstructural differences play an important role in the behavior of materials. In fact, it has been shown that the presence of undissolved phases in 7000 series alloys has a detrimental effect on some important properties;[3,36-39] iii) to have materials with different grain structures. For this reason, sheet, plates, and extrusions were considered. Typical microstructures of the wrought products in the fully heat treated condition are shown in Fig. 3. It can be seen that the sheets have a completely recrystallized structure, the plates have a partially recrystallized structure, *i.e.* subgrain structure, and the extrusions have a fine subgrain structure. It can also be seen that the high purity alloys have a much lower volume percent of undissolved second phase particles than do the commercial purity alloys.

b) Cycle Selection

The cycles were selected so that the effect of the deformation (H) in relation to the artificial aging stage (A) could be investigated.

The cycles used were T-A, T-AA, T-HA, T-HAA,

Table I. Materials Investigated

Alloy	No.	Zn Pct	Mg Pct	Cu Pct	Cr Pct	Mn Pct	Zr Pct	Ti Pct	Fe Pct	Si Pct	Wrought Products
7005 commercial purity	1	5.07	1.04	–	0.14	0.22	0.12	–	0.12	0.09	4 mm-thick sheets and 70 by 7 mm extruded bars produced from an industrial dc cast ingot
7039 commercial purity	2	4.01	2.78	–	0.14	0.23	0.10	-	0.14	0.09	4 mm-thick sheets produced from an industrial dc cast ingot
7039 high purity	3	4.09	3.04	–	0.18	0.33	0.14	0.045	0.012	0.017	2 mm-thick sheets and 15 mm-diam extruded bars produced from laboratory dc cast 110 mm-diam ingot
7075 commercial purity	4	5.49	2.46	1.69	0.13	0.20	–	0.074	0.28	0.20	2 mm and 5 mm-thick sheets and 15 mm-diam extruded bars produced from an industrial dc cast ingot
7075 commercial purity	5	5.78	2.10	1.68	0.19	0.05	–	0.057	0.12	0.056	35 mm-thick industrial plate
7075 commercial purity	6	5.74	2.60	1.63	0.22	-	–	0.005	0.22	0.13	2 mm-thick sheets, and 50 by 12 mm extruded bars produced from laboratory dc cast 110 mm-diam ingot
7075 high purity	7	5.68	2.50	1.60	0.22	–	–	0.010	0.0016	0.005	2 mm-thick sheets, 15 mm-diam and 50 by 12 mm extruded bars produced from laboratory dc cast 110 mm-diam ingot

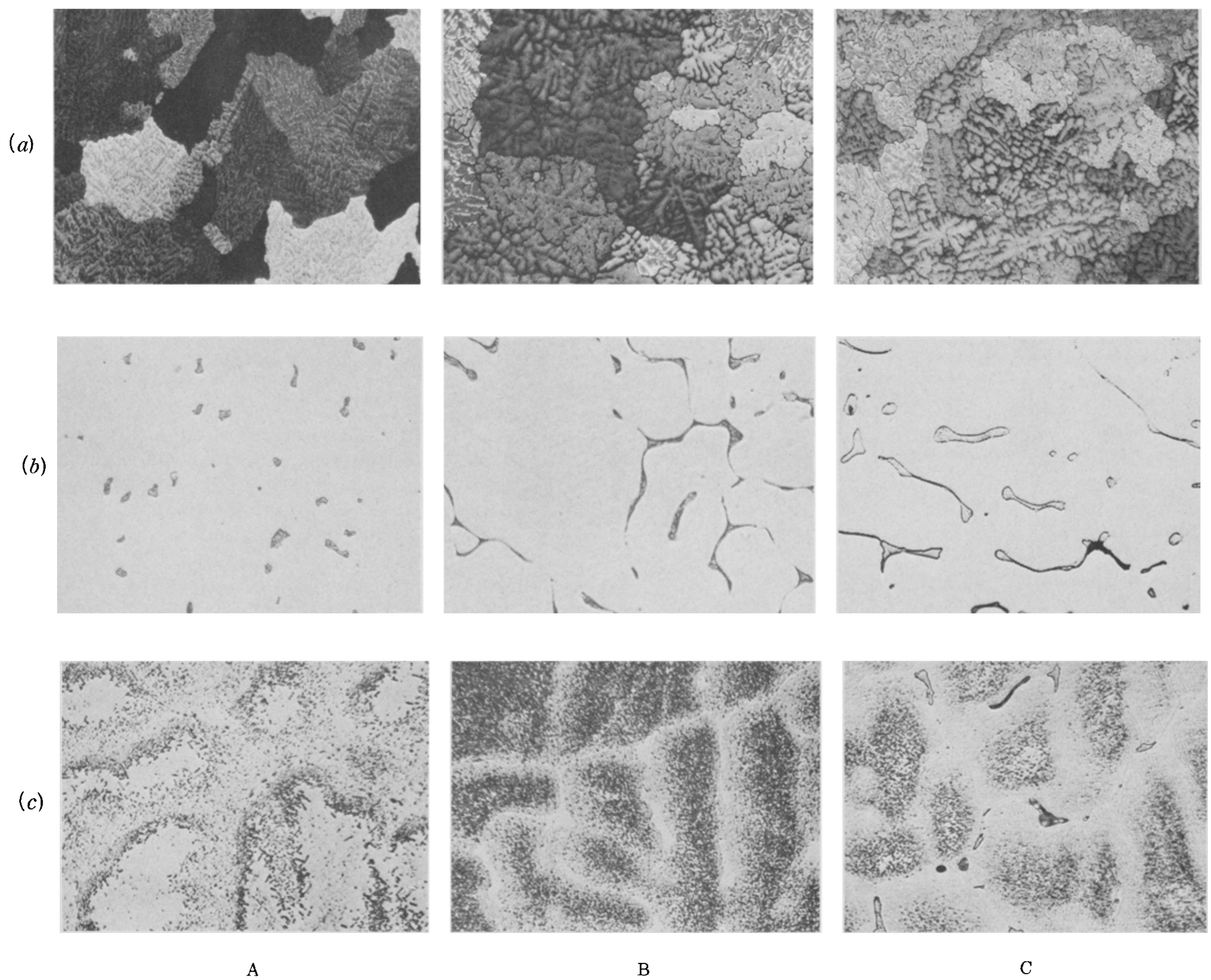


Fig. 2—Typical microstructures of the dc cast 110 mm-diam ingots of the 7039 h.p. (A), 7075 h.p. (B), and 7075 c.p. (C) alloys: (*a*) as cast: magnification 50 times; electrolytic etching in 2 pct HF; (*b*) as cast: magnification 475 times; etching in 0.5 pct HF; (*c*) homogenized at 450°C for 8 h + 480°C for 15 h; magnification 475 times; etching in 0.5 pct HF.

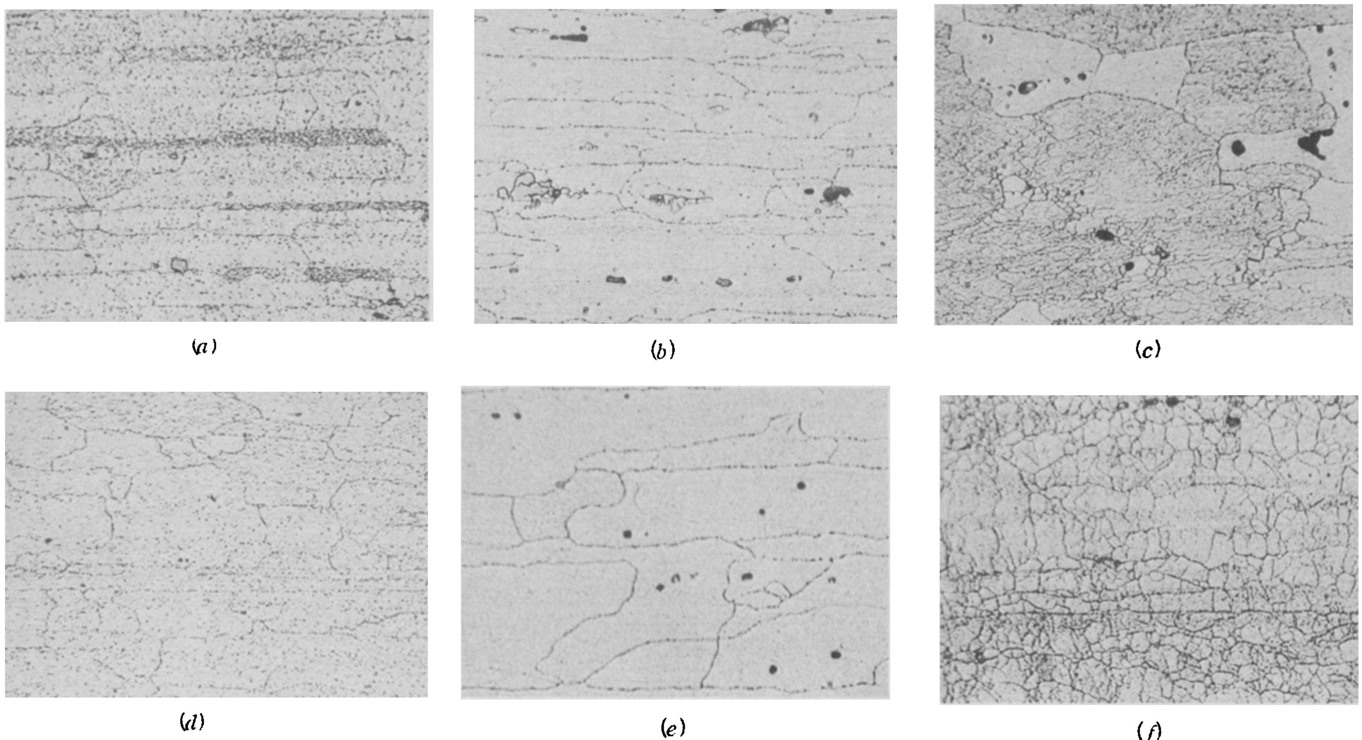


Fig. 3—Typical microstructures of some wrought products of 7039 and 7075 alloys, after full heat treatment (T6) observed on the longitudinal section. Magnification 475 times; etching: 7039 → in 55 pct H_3PO_4 at 50°C; 7075 → in 25 pct HNO_3 at ~ 70°C. (*a*) 7039 c.p., 4 mm-thick sheet; (*b*) 7075 c.p., 5 mm-thick sheet; (*c*) 7075 c.p., 35 mm-thick plate; (*d*) 7039 h.p., 2 mm-thick sheet; (*e*) 7075 h.p., 2 mm-thick sheet; (*f*) 7075 c.p., 16 mm-diam extrusion.

T-AH, T-AAH, T-AHA, and many combinations of A and H were selected. The values of H ranged from 5 to 50 pct reduction in thickness for sheets and from 5 to 20 pct reduction in thickness for plates and extrusions. The deformation was applied by rolling or drawing[31] both at room temperature and in the range 130° to 220°C with preheating times ranging from 30 s up to 20 min. The first aging step, on the basis of hardness tests,[40,41] was carried out at 100°C for 10 h for the ternary alloys and at 105°C for 6 h for the quaternary alloy.

c) Testing

All specimens were tested in tension in the longitudinal direction and for the 35 mm thick plate (alloy No. 5) and the 50 by 12 mm extrusions (alloy Nos. 6 and 7) also in the long transverse direction. At least two specimens were tested for each treatment.

The fracture toughness tests were carried out on 50 by 12 mm extruded laboratory produced 7075 (alloy Nos. 6 and 7) in the long transverse direction.

The specimen used to determine K_c had a single side notch and was similar to the specimen proposed by Sullivan.[42,43] All of the values of K_c reported here are the average of 3 measurements.

The stress corrosion tests were performed on 35 mm thick plate of commercial purity 7075 (alloy No. 5) in the short transverse direction. "C" ring specimens were taken from the plate in the T-A, T-AA, and T-AHA tempers. The specimens were loaded to 50 and 75 pct of the yield strength and tested in alternate immersion in a 3.5 pct NaCl solution. Each cycle consisted of 1 h (10 min immersion and 50 min exposure to air) and was carried out in a thermostatically controlled chamber at 20° ± 1°C. The test ended when cracks became visible. In the cases where no cracks were visible after 300 h, the samples were removed after various times up to a limit of 1000 h, unloaded and examined for the presence of plastic deformation.

If plastic deformation had occurred, optical microscopy was used to determine if stress corrosion cracks were present. Also, in doubtful cases and always after 1000 h testing, optical microscopy was used. At least 5 samples were tested for each temper in order to determine the specific time or the time interval where stress corrosion cracking had occurred.

The fatigue tests were carried out at room temperature on cantilever beam specimens using a rotating beam testing machine operating at 11,500 rpm. Both smooth and notched specimens (Kt = 4.8) of 15 mm diam extruded bars of high purity 7039 and 7075 (alloy Nos. 3 and 7) were tested in the T-AA (ternary alloy), T-A (quaternary alloy), and T-AHA tempers. Also commercial purity 7075 (alloy No. 4) was tested in the T-A temper. The metallographic aspects of the materials were observed by electron microscopy, to establish the relationships between properties and structure.

3) RESULTS AND DISCUSSION

a) Tensile Properties

The tensile properties of the most interesting treatments are given in Tables II to VI. It can be seen that with FTMT (T-AHA cycles) all the materials tested had significant increases in UTS and YS, with an acceptable decrease in elongation, compared to the values obtained with traditional cycles.

Table II. Longitudinal Tensile Properties Shown by 4 mm-Thick Sheets of 7005 and 7039 Alloys of Industrial Production in Several Tempers

Alloy	Aging Treatment					Tensile Properties		
		Deformation						
	1st Step	Type	Temperature	Deg Pct	2nd Step	UTS, psi	YS, psi	E, Pct
	100°C for 200 h	–	–	–	–	57,000	49,500	16.5
	100°C for 10 h	–	–	–	130°C for 22 h	55,000	48,000	17.0
	100°C for 10 h	rolling	r.t.	20	100°C for 100 h	63,000	59,500	12.4
7005	100°C for 10 h	rolling	r.t.	20	115°C for 15 h	63,000	59,000	11.5
commercial purity	100°C for 10 h	rolling	r.t.	20	130°C for 8 h	62,000	57,000	11.1
No. 1	100°C for 10 h	rolling	r.t.	40	100°C for 50 h	67,000	62,500	8.7
	100°C for 10 h	rolling	r.t.	40	115°C for 10 h	65,500	61,000	8.8
	100°C for 10 h	rolling	r.t.	40	130°C for 4 h	64,000	60,000	8.6
	100°C for 200 h	rolling	r.t.	20	–	67,000	65,000	5.5
	100°C for 200 h	rolling	r.t.	40	–	71,000	69,000	4.9
	100°C for 200 h	–	–	–	–	65,000	55,500	16.0
	100°C for 10 h	–	–	–	130°C for 22 h	63,500	55,000	15.4
	100°C for 10 h	rolling	r.t.	20	100°C for 60 h	74,000	69,500	10.2
7039	100°C for 10 h	rolling	r.t.	20	115°C for 15 h	72,500	67,000	11.2
commercial purity	100°C for 10 h	rolling	r.t.	20	130°C for 6 h	71,500	66,000	9.8
No. 2	100°C for 10 h	rolling	r.t.	20	100°C for 30 h	77,000	71,000	8.1
	100°C for 10 h	rolling	r.t.	20	115°C for 10 h	76,000	70,000	7.8
	100°C for 10 h	rolling	r.t.	20	130°C for 4 h	74,000	69,000	8.0
	100°C for 200 h	rolling	r.t.	20	–	76,500	74,000	5.6
	100°C for 200 h	rolling	r.t.	40	–	82,000	80,000	5.0

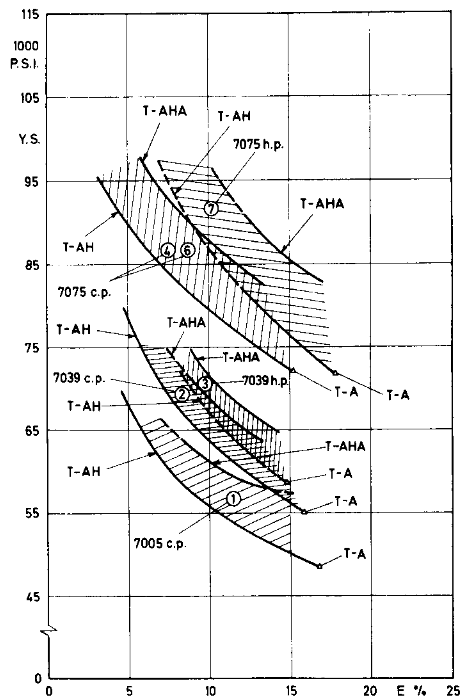


Fig. 4—Yield strength values vs elongation of sheets of 7005, 7039, and 7075 alloys in several thermal and thermomechanical tempers.

Sheet exhibits the best combination of strength and ductility, with increases in UTS and YS from 15 to 30 pct and decreases in elongation of 2 to 8 points, equivalent to 10 to 50 pct compared to the traditional tempers T-A and to processes of basic T-AH type. In Fig. 4 the shaded areas represent the range of values obtained with T-AHA cycles and the lower curve is defined by the values relative to the T-A, T-AA, T-AH, and T-AAH tempers for each alloy.

Less of an increase in strength was found in plate and especially in extrusions, which, because of their micrograins and oriented structure, have high tensile properties even in the traditional tempers. It should be pointed out that other treatments comprising strain and aging do not give favorable strength-ductility combinations.

For example, the T-AH or T-AAH cycles, at a level of strength equivalent to that obtained by the T-AHA cycle, show a loss of elongation from the T6 temper of 10 to 12 points which is equivalent to a reduction of 70 to 80 pct.

If the deformation is applied before artificial aging (T-HA cycles) the peak strength values are lower, other conditions being equal, than those obtained with the T-AHA cycle, Table V. Also, if H is not very high, the final peak values are lower than those obtained with conventional thermal treatments. The intermediate deformation increases the aging rate during the final artificial aging step; for example, in 4 mm thick sheet of 7005, after an initial aging at 100°C to 10 h followed by 20 pct deformation, the peak strength during subsequent aging at 100°C is reached in about 100 h, whereas in the normal isothermal aging treatment the peak strength does not occur before 200 h, Table II. It has been found that the deformation should not exceed 30 pct for sheets, 20 pct for plates, and 10 pct for extrusions since higher degrees of deformation cause only a slight increase in strength accompanied, however, by a large decrease in elongation.

Regarding warm deformation, it can be applied with no deleterious effect in the range 130° to 170°C for the ternary alloys (preheating times 10 to 20 min) and in the range 170° to 220°C for the quaternary alloys (preheating times 2.5 to 15 min up to 190°C and 30 s to 2.5 min for higher temperatures). As can be seen for 7039 alloy, Table III, for equal final properties, the strengthening achieved by the last aging step is higher in the case of warm deformation than for the cycle with H applied at room temperature. For 7075, the use of warm deformation, in addition to producing increased plasticity of the alloy, results in a better combination

Table III. Longitudinal Tensile Properties of 4 mm-Thick Sheets of 7039 C.P. Alloy (No. 2) in Several Thermomechanical Tempers, with Deformation Given at Various Temperatures

Aging Treatment					Tensile Properties		
	Deformation						
1st Step	Type	Temperature	Deg Pct	2nd Step	UTS, psi	YS, psi	E, Pct
100°C for 10 h	–	–	–	130°C for 22 h	63,500	55,000	15.4
100°C for 10 h	–	–	–	–	54,000	30,000	22.9
100°C for 10 h	rolling	r.t.	20	–	66,000	58,000	9.0
100°C for 10 h	rolling	r.t.	20	115°C for 15 h	72,000	66,500	10.2
100°C for 10 h	rolling	130°C	20	–	64,500	55,000	10.0
100°C for 10 h	rolling	130°C	20	115°C for 15 h	72,000	66,500	10.0
100°C for 10 h	rolling	150°C	20	–	62,500	55,000	10.5
100°C for 10 h	rolling	150°C	20	115°C for 15 h	71,500	65,500	9.7
100°C for 10 h	rolling	160°C	20	–	59,500	52,000	9.5
100°C for 10 h	rolling	160°C	20	115°C for 15 h	70,000	65,000	10.1
100°C for 10 h	rolling	170°C	20	–	61,000	52,000	9.7
100°C for 10 h	rolling	170°C	20	115°C for 15 h	69,500	62,500	9.4

Table IV. Tensile Properties of Sheets (Longitudinal Direction) and Plates (Long Transverse Direction) of 7075 C.P. Alloys of Industrial Production in Several Tempers

Material		Aging Treatment					Tensile Properties		
			Deformation						
Alloy	Wrought Product	1st Step	Type	Temperature	Deg Pct	2nd Step	UTS, psi	YS, psi	E, Pct
7075, No. 4	5 mm-thick sheet	120°C for 24 h	–	–	–	–	83,000	73,500	13.1
		105°C for 6 h	rolling	r.t.	20	105°C for 50 h	95,500	91,500	7.8
		105°C for 6 h	rolling	r.t.	20	120°C for 8 h	92,000	88,000	8.0
		105°C for 6 h	rolling	r.t.	50	120°C for 4 h	98,000	94,000	4.8
		120°C for 24 h	rolling	r.t.	20	–	95,500	92,000	4.0
		105°C for 6 h	rolling	160°C	20	120°C for 8 h	92,000	86,000	7.7
		105°C for 6 h	rolling	190°C	20	120°C for 8 h	91,000	86,500	8.2
	2 mm-thick sheet	120°C for 24 h	–	–	–	–	81,000	71,000	16.7
		105°C for 6 h	rolling	r.t.	10	120°C for 13 h	89,000	83,000	13.0
		105°C for 6 h	rolling	r.t.	20	120°C for 7 h	92,500	88,000	10.1
		105°C for 6 h	rolling	170°C	10	120°C for 13 h	88,500	86,000	12.3
		105°C for 6 h	rolling	170°C	20	120°C for 6 h	93,000	88,000	10.5
		105°C for 6 h	rolling	190°C	10	120°C for 13 h	90,000	85,000	11.7
		105°C for 6 h	rolling	190°C	20	120°C for 5 h	92,000	89,500	10.7
7075, No. 5	35 mm-thick plate	120°C for 24 h	–	–	–	–	78,500	71,000	8.0
		105°C for 6 h	rolling	175°C	10	120°C for 13 h	83,000	78,500	5.0
		105°C for 6 h	rolling	180°C	20	120°C for 7 h	83,000	79,500	5.0

of tensile properties than is obtained by the use of room temperature deformation, Fig. 5 and Table IV.

In all of the T-AHA cycles studied the high purity 7039 and 7075 sheets have good toughness even though the notch toughness slightly decreases with increasing degrees of deformation. On the other hand, in commerical purity 7075, greater than 10 pct deformation caused a significant decrease in notch toughness, Table V. The effect of FTMT on the strength of both commercial purity and high purity alloys is the same. It has been found that high purity 7075 sheet has higher ductility, and especially, notch toughness, than its commercial purity counterpart, independent of the type of treatment used, Table V. In practical terms, this means that by using high purity alloys, materials produced in T-AHA temper at 15 pct higher yield strength have equal or higher elongation and reduction in area values, higher notch strength and notch toughness as compared to the corresponding commercial purity alloys in the T-A temper. It should also be pointed out, Tables V and VI, that there are some differences in the tensile properties of the industrially produced and laboratory produced commercial purity alloys.

Generally, the alloys prepared in the laboratory have better combinations of strength and ductility for a given temper; this is probably due to their smaller dendrite arm spacings.

b) Fracture Toughness

The results of fracture toughness tests along with the corresponding tensile properties of samples of laboratory produced 7075 in the T-A and T-AHA tempers are shown in Table VII. It can be seen that although FTMT generally causes a decrease in toughness, h.p. 7075 in the T-AHA state (H = 10 pct) does have comparable elongation values, 17 pct higher K values and 12 pct higher yield strength values than commercial purity 7075 in the T-A temper.

c) Stress Corrosion Resistance

The short-transverse stress corrosion resistance and the corresponding long transverse tensile prop-

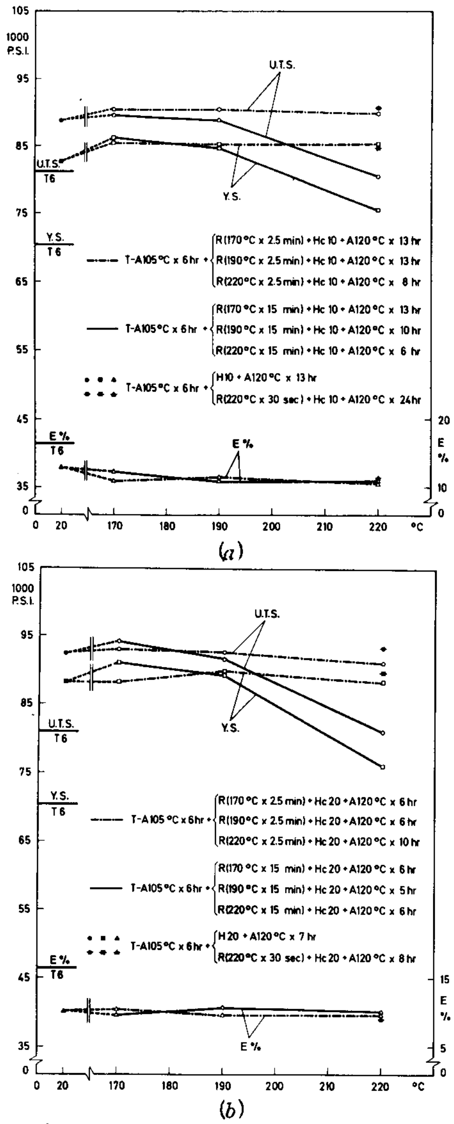


Fig. 5—Tensile properties of 2 mm-thick sheet of 7075 c.p. alloy (No. 4) subjected to thermomechanical treatments of T-AHA type, with H = 10 pct (*a*) and 20 pct (*b*), given at different temperatures. The times of final aging at 120°C correspond to peak strength.

erties of the 35 mm thick plates of commercial purity 7075 in various T-AHA cycles, as well as the T6 (T-A), T73 (T-AA), and a stress corrosion resistant T-AA temper,[44] are presented in Table VIII.

It can be seen from this table that T-AHA cycles with the second aging step at low temperatures (specimens *d* and *e* in Table VIII) have higher strength than the T6 material but do not show any improvement in stress corrosion resistance.

In the T-AHA cycle where the second aging step is carried out at a higher temperature (specimen *f* in Table VIII), at strength levels equal to or slightly higher than those of the T6 temper, there is a marked improvement in the stress corrosion resistance over the T6 temper. However, it does appear that the complex dislocation-precipitation structure obtained with FTMT has only a small effect on the stress corrosion resistance since in the T-AA and T-AHA states which have the same aging treatments (specimens *c* and *f* in Table VIII), the stress corrosion resistance is the same.

d) Fatigue

The tensile properties of the 7039 and 7075 tested are shown in Table VI. The S-N curves for the fatigue tests on the 15 mm diam extruded bars of high purity 7039 and commercial and high purity 7075 are shown in Figs. 6 and 7, respectively. The S-N curve of smooth specimens of 7039 T-AA does not show a well defined limit while that of the T-AHA material has a higher slope up to about 10^7 cycles and has a well defined asymptote. The fatigue strength at 5×10^8 cycles for the materials in the T-AA and T-AHA tempers are 30,000 psi and 28,500 psi, respectively.

In the notched specimens of 7039, the T-AA and T-AHA tempers show fatigue strengths which at 5×10^8 cycles are 14,000 psi and 9,000 psi, respectively. As in the case of the smooth samples, material in the T-AHA temper has a higher slope at short times and a more pronounced limit which becomes stable at a lower number of cycles. The effective notch coefficients (K eff) which is the ratio between the limit (or the fatigue strength at 10^8 cycles) for the smooth specimens and that for the notched ones are: T-AA temper K eff = 2.10 and T-AHA K eff = 3.33. Thus, it appears that T-AHA makes the alloy more notch sensitive than the T-AA temper.

In smooth specimens of commercial purity 7075 there is no definite limit observed up to 10^9 cycles. The high purity 7075 in the T-A temper has a well defined limit after about 10^7 cycles and has a fatigue strength at 5×10^8 cycles of 35,000 psi. This value is higher than the fatigue strength of the commercial purity material at 5×10^8 cycles. In the T-AHA material the S-N curve is regular up to about 5×10^7 cycles and with increased cycles, there is a sharp decrease in the fatigue strength with a discontinuity phenomenon, or "knee" effect[45,46] which is probably associated with differences in the

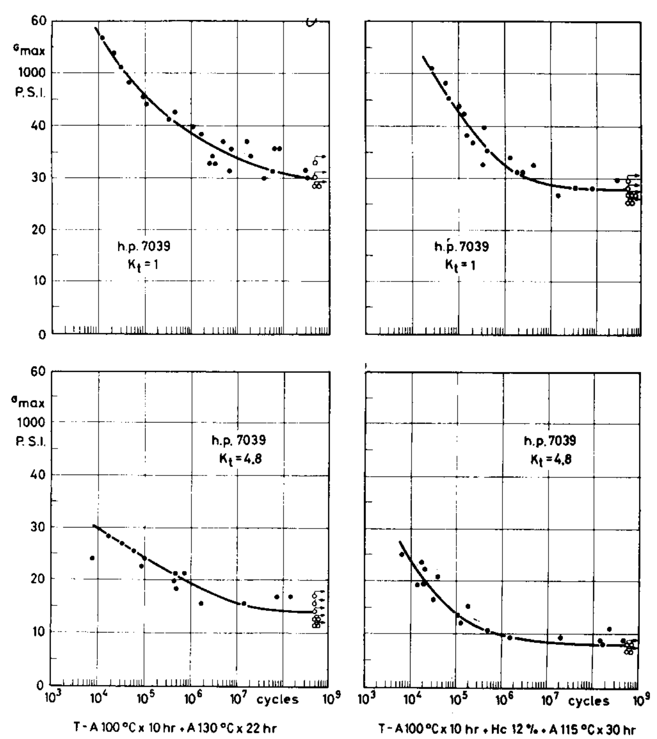


Fig. 6—S-N curves for 15 mm-diam extruded rods of 7039 h.p. alloy (No. 3) with various treatments, obtained on smooth (k_t = 1) and notched (k_t = 4.8) specimens.

Table V. Longitudinal Tensile Properties (Smooth and Notched (k_t = 10.2) Specimens) of 2 mm-Thick Sheets of 7039 and 7075 Laboratory Produced Alloys, in Several Tempers

Alloy	Aging Treatment: 1st Step	Deformation: Type	Deformation: Temperature	Deformation: Deg Pct	2nd Step	Tensile Properties: UTS, psi	YS, psi	E, Pct	RA, Pct	NTS, psi	NTS, YS
7075 high purity No. 7	120°C for 24 h	–	–	–	–	84,500	72,000	17.8	26.9	86,500	1.20
	105°C for 6 h	rolling	175°C	10	120°C for 13 h	90,000	84,000	16.2	27.2	94,500	1.12
	105°C for 6 h	rolling	175°C	20	120°C for 6 h	94,000	89,000	13.0	19.4	92,400	1.04
	–	rolling	r.t.	10	120°C for 24 h	82,000	69,000	16.0	–	–	–
7075 commercial purity No. 6	120°C for 24 h	–	–	–	–	87,000	75,000	15.3	28.0	81,000	1.08
	105°C for 6 h	rolling	175°C	10	120°C for 13 h	93,000	86,500	10.4	22.9	79,000	0.91
	105°C for 6 h	rolling	175°C	20	120°C for 6 h	95,500	91,000	10.0	16.8	75,000	0.82
7039 high purity No. 3	100°C for 10 h	–	–	–	130°C for 22 h	66,000	58,000	14.8	38.5	70,500	1.21
	100°C for 10 h	rolling	155°C	10	115°C for 28 h	72,000	67,000	12.4	32.4	76,000	1.14
	100°C for 10 h	rolling	155°C	20	115°C for 15 h	72,500	68,500	10.6	30.9	78,000	1.14
	100°C for 10 h	rolling	155°C	40	115°C for 9 h	75,500	72,000	10.5	18.6	79,000	1.10

Table VI. Tensile Properties of 15 mm-Diam Extruded Rods of 7039 and 7075 Alloys of High and Commercial Purity in Several Tempers

Alloy	Aging Treatment: 1st Step	Deformation: Type	Deformation: Temperature	Deformation: Deg Pct	2nd Step	Tensile Properties: UTS, psi	YS, psi	E, Pct
7039 commercial purity, No. 2	100°C for 10 h	–	–	–	130°C for 22 h	82,000	77,000	8.2
	100°C for 10 h	drawing	r.t.	20	115°C for 15 h	87,000	85,500	4.8
7039 high purity, No. 3	100°C for 10 h	–	–	–	130°C for 22 h	83,500	77,000	7.5
	100°C for 10 h	drawing	155°C	12	115°C for 30 h	85,000	82,000	6.7
7075 commerical purity, No. 4	120°C for 24 h	–	–	–	–	101,000	94,000	5.7
	105°C for 6 h	drawing	180°C	10	120°C for 10 h	106,000	105,000	3.4
7075 high purity No. 7	120°C for 24 h	–	–	–	120°C for 24 h	100,500	96,500	6.2
	105°C for 6 h	drawing	175°C	5.5	120°C for 24 h	104,000	101,500	5.6
	105°C for 6 h	drawing	175°C	12	120°C for 13 h	105,000	103,000	4.3

Table VII. Long-Transverse Tensile and Fracture Toughness Properties of 50 by 12 mm Extruded Bar of 7075 Laboratory Produced Alloys, in Several Tempers

Alloy	Aging Treatment: 1st Step	Deformation: Type	Deformation: Temperature	Deformation: Deg Pct	2nd Step	Tensile Properties: UTS, psi	YS, pis	E, Pct	Fracture Toughness Index, psi $\frac{Kc}{\sqrt{in.}}$
7075 commercial purity No. 6	120°C for 24 h	–	–	–	–	87,000	78,500	8.8	40,500
	105°C for 6 h	rolling	175°C	10	120°C for 13 h	93,000	86,000	5.6	28,000
	105°C for 6 h	rolling	175°C	20	120°C for 6 h	92,500	89,000	5.0	27,000
7075 high purity No. 7	120°C for 24 h	–	–	–	–	89,000	81,500	10.6	57,000
	105°C for 6 h	rolling	175°C	10	120°C for 13 h	93,500	87,500	8.1	47,000
	105°C for 6 h	rolling	175°C	20	120°C for 6 h	95,000	89,500	7.8	37,500

mechanism of either crack initiation or crack propagation at different stress levels. For this reason it is difficult to define a fatigue limit and thus, one can only describe a length of time at a certain load level. In this sense, the alloy does not fail up to about 5×10^8 cycles under a stress of about 25,000 psi.

The notched specimens of 7075 in the T-A temper show a typical curve with the "knee" effect and a well defined fatigue limit at 10^7 cycles of about 11,000 psi. This limit appears to be independent of the degree of purity of the alloy. The experimental values for the S-N curve of notched specimens of 7075 in the T-AHA temper follow two curves which appear to be sufficiently different to suggest that this behavior is not due to usual scatter in the data. Up to 10^8 cycles, the upper curve is quite different from the T-A material, while the lower curve is typical for notched specimens with a well defined limit at 10^6 cycles. The effective notch coefficients for 7075 in the T-A temper and the T-AHA temper are 3.22 and 1.77, respectively. These values show that there is a marked reduction of the notch effect in the samples produced with FTMT cycles, showing quite a different behavior of 7075 from 7039.

The results on the whole suggest that T-AHA cycles produce a decrease in the fatigue resistance of the ternary alloys, in particular in notched specimens. In the case of the quaternary alloy, T-AHA cycles lead to more reduced fatigue properties on smooth specimens,

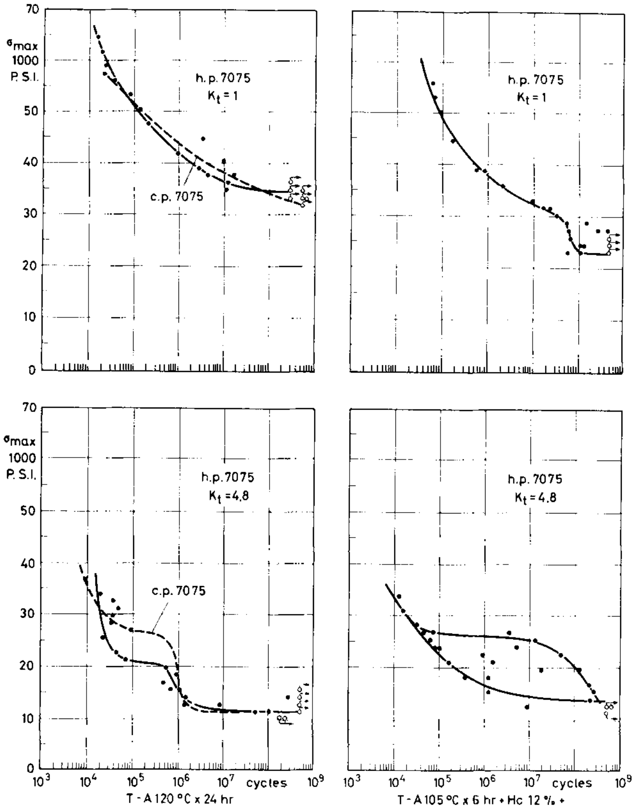


Fig. 7—S-N curves for 15 mm-diam extruded rods of 7075 h.p. alloy (No. 7) with various treatments, obtained on smooth ($k_t = 1$) and notched ($k_t = 4.8$) specimens. For comparison S/N curves for similar samples of 7075 c.p. alloy (No. 4-dotted line) in T-A120°C for 24 h are reported.

while causing an improvement in notched samples. However, the discrepancies in the data do not permit unambiguous interpretation especially since, in respect to other metallurgical properties, the relationship between the fatigue processes and microstructure is not well known. Thus it appears a difficult task to explain the S-N curves in terms of dislocations and precipitation distributions.

e) Transmission Electron Microscopy

Transmission electron microscopy examinations show that, similarly to what has been reported elsewhere for the ternary alloys,[23] the improvement in strength produced by FTMT with respect both to conventional thermal treatments and to other processes based on different combinations of H and A, can be attributed to a better hardening type structure in the FTMT samples.

Specifically, Figs. 8 to 10 show at low magnification that high purity 7075 sheets have uniform dislocation structure in the T-H10 pct A120°C for 24 h and T-A 105°C for 6 h H10 pct A120°C for 13 h states in contrast to the defect free matrix in the T-A sample. At higher magnification, the T-HA sample, aged to peak hardness, has a coarse precipitation of η' (or η) particles with respect to a fine distribution in both the T-A and T-AHA cycles, Figs. 11 to 13. In other words, it appears that T-AHA samples have the optimum combination of dislocations and precipitate structure for effective strengthening.

A typical feature of the T-AHA samples is the presence of wide slip bands overlapping the uniform dis-

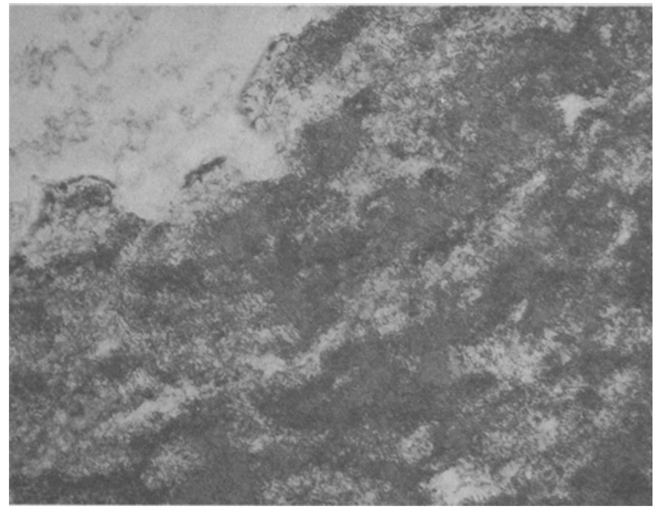


Fig. 8—Transmission electron micrograph of 2 mm-thick sheet of 7075 h.p. alloy (No. 7) in T-H10 pct A120°C for 24 h temper. Magnification 19,200 times.

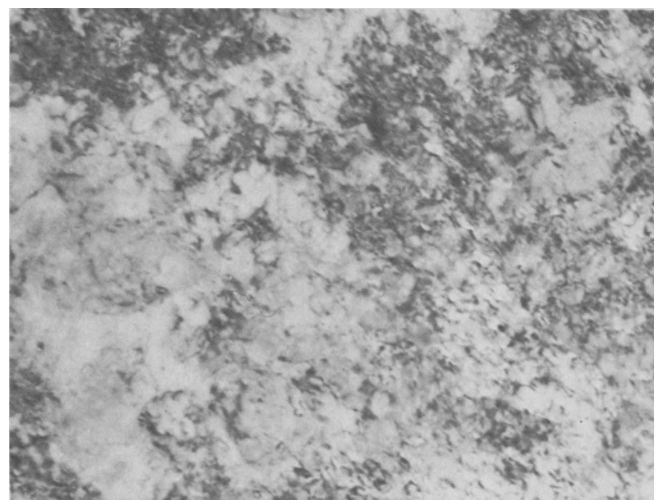


Fig. 9—Transmission electron micrograph of 2 mm-thick sheet of 7075 h.p. alloy (No. 7) in T-A105°C for 6 h H10 pct A120°C for 13 h temper. Magnification 48,000 times.

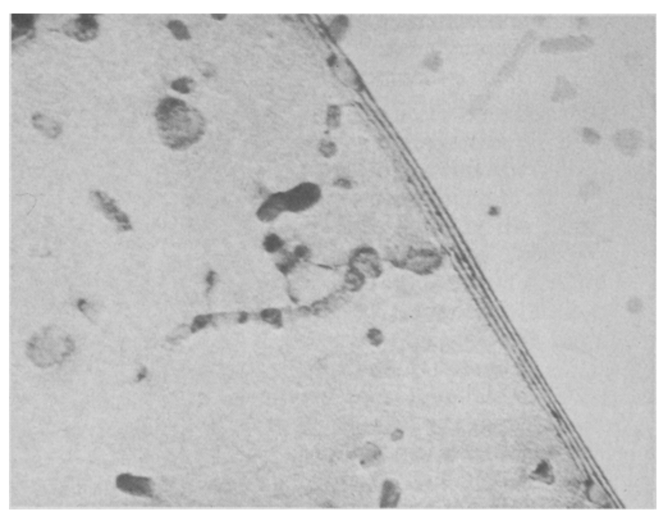


Fig. 10—Transmission electron micrograph of 2 mm-thick sheet of 7075 h.p. alloy (No. 7) in T-A120°C for 24 h temper. Magnification 47,500 times.

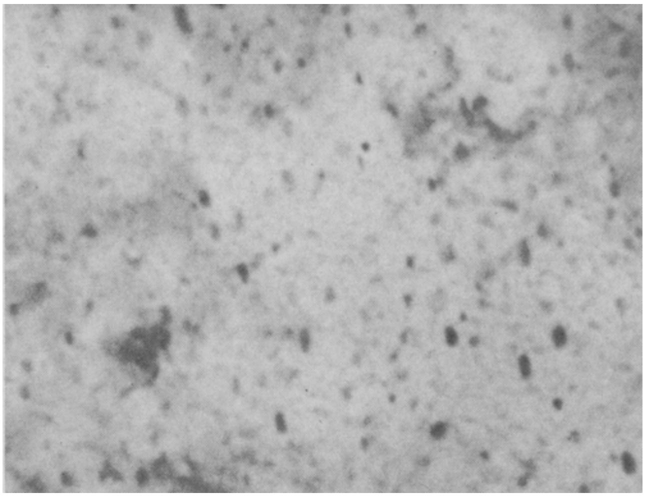


Fig. 11—Transmission electron micrograph of 2 mm-thick sheet of 7075 h.p. alloy (No. 7) in T-H10 pct A120°C for 24 h temper. Magnification 190,000 times.

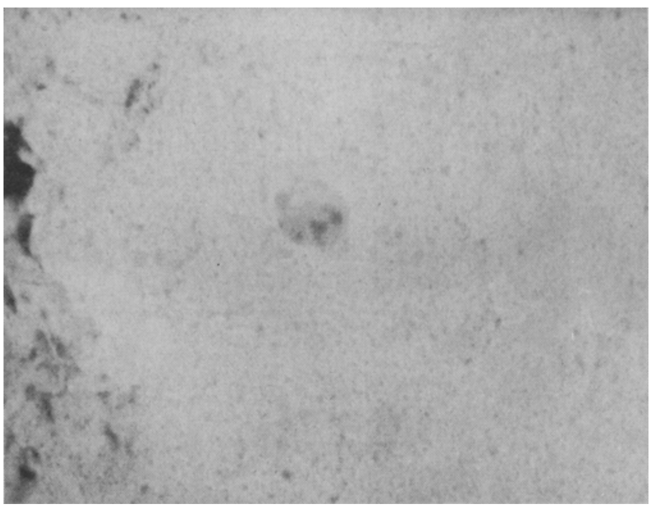


Fig. 12—Transmission electron micrograph of 2 mm-thick sheet of 7075 h.p. alloy (No. 7) in T-A105°C for 6 h H10 pct A120°C for 13 h temper. Magnification 192,000 times.

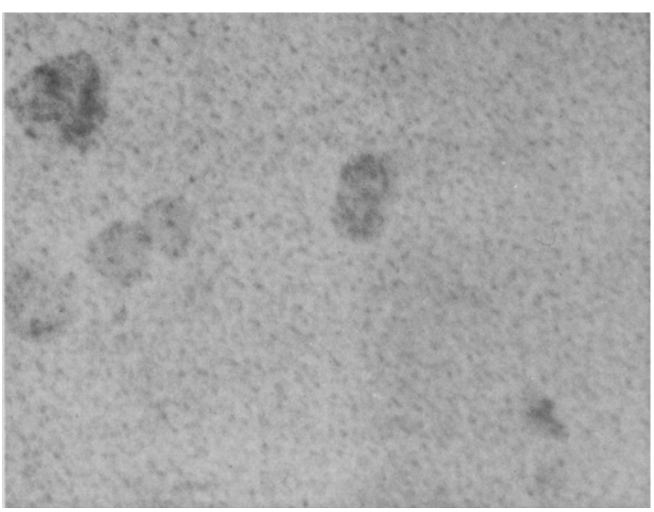
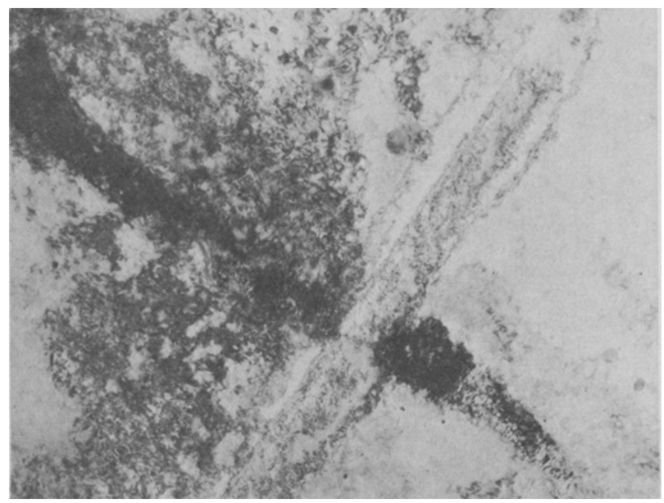


Fig. 13—Transmission electron micrograph of 2 mm-thick sheet of 7075 h.p. alloy (No. 7) in T-A120°C for 24 h temper. Magnification 192,000 times.

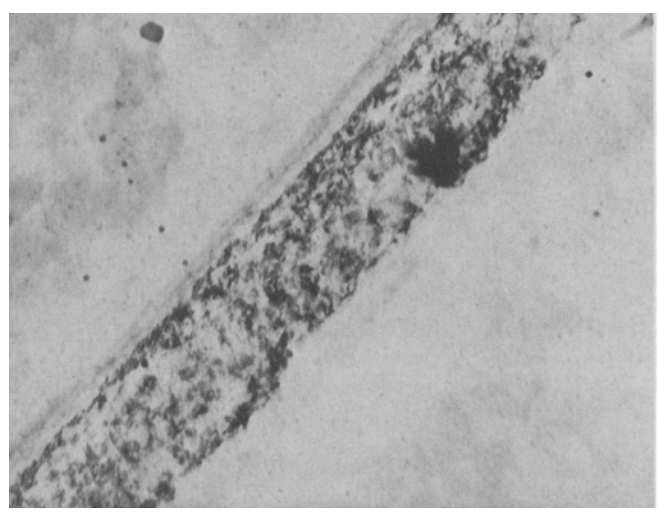
location structure and characterized by dense tangles of dislocations hardly resolvable also at high magnifications, Fig. 14(*a*),(*b*). The tendency to form such slip bands has been found to be greater in samples which have a high degree of intermediate deformation and in wrought products, such as extrusions, which are constituted by a subgrain structure. The presence of these coarse slip bands may play a role in the loss of ductility in FTMT material.

4) CONCLUSIONS

The increase in strength without drastic loss in ductility achieved in medium and high strength Al-Zn-Mg(-Cu) alloys by the application of T-AHA type FTMT is quite significant. This increase in strength is probably due to a superposition of the strengthening effects from dislocations and from precipitation. The most beneficial effect of FTMT occurs in high purity recrystallized wrought products, *i.e.* sheet; in these materials, the YS and UTS can be increased by 20 pct with the same or better ductility and notch and fracture toughness compared to commercial



(*a*)



(*b*)

Fig. 14—Typical aspect of coarse slip bands observed on 4 mm-thick sheets of 7005 c.p. alloy (No. 1), treated according to a T-AHA cycle. Magnification (*a*) 16,800 times; (*b*) 48,000 times.

Table VIII. Tensile (Long Transverse) and Stress-Corrosion (Short Transverse) Properties Determined on 35 mm-Thick Plate of 7075 C.P. Alloy (No. 5) of Industrial Production, in Several Tempers

	Aging Treatment					Tensile Properties			Stress Corrosion Properties		
		Deformation							Applied Load		
Type	1st Step	Type	Temp.	Deg Pct	2nd Step	UTS, psi	YS, psi	E, Pct	Pct YS	Time of Test, h	Optical Microscopy Observation
a)	120°C for 24 h	–	–	–	–	78,500	71,000	8.0	75	48 to 72	Intergranular cracks (sometimes passing through the thickness) which start from the exposed tensioned surface
									50	160 to 300	Dispersed shallow pits (~0.15 mm); intergranular cracks (not passing through) starting from the bottom of some pits or directly from the exposed tensioned surface
b)	105°C for 6 h	–	–	–	170°C for 8 h	75,000	67,500	7.7	75	≃1000	Dispersed pits rather deep (~0.30 mm); intergranular cracks (not passing through) starting from the bottom of some pits. Some transgranular cracks in the corroded zones
									50		–
c)	105°C for 6 h	–	–	–	150°C for 8 h	79,000	72,000	8.0	75	350 to 500	Dispersed pits more or less deep (~0.25 mm); intergranular cracks (not passing through) starting from the bottom of some pits or directly from the exposed tensioned surface
									50	800 to 1000	Dispersed pits rather deep (~0.35 mm); intergranular cracks (not passing through) starting from the bottom of some pits
d)	105°C for 6 h	rolling	175°C	10	120°C for 13 h	83,000	78,500	5.0	75	40 to 80	Intergranular cracks (sometimes passing through) starting from the exposed tensioned surface
									50	100 to 250	Some shallow dispersed pits (~0.1 mm); intergranular cracks (sometime passing through) starting in general directly from the exposed tensioned surface
e)	105°C for 6 h	rolling	180°C	20	120°C for 7 h	83,000	79,500	5.0	75	≃24	Intergranular cracks (sometimes passing through) starting from the exposed tensioned surface
									50	40 to 90	
f)	105°C for 6 h	rolling	180°C	10	150°C for 8 h	79,000	73,500	7.9	75	400 to 500	Dispersed pits more or less deep (~0.25 mm); intergranular cracks (not passing through) starting from the bottom of some pits or directly from the exposed tensioned surface
									50	800 to 1000	Dispersed pits rather deep (~0.35 mm); intergranular cracks (not passing through) starting from the bottom of some pits

conventionally aged material. FTMT does not produce significant variation in the typical stress corrosion resistance but may cause a slight decrease in the fatigue resistance; however, this work does show that FTMT in combination with the use of high purity alloys does give a method of producing materials with much higher strength and the same or better secondary properties than conventionally available materials.

It should be pointed out that there are no problems in the application of FTMT cycles to simple shaped wrought products, especially sheet and plate. The working operations can be carried out at relatively high temperatures, so that the application of the deformation itself is facilitated. Also, there are no close tolerance limitations on the deformation temperatures or on the aging treatments, so that FTMT cycles seem to be quite compatible with good production control standards.

ACKNOWLEDGMENTS

The authors acknowledge the valuable assistance provided by Mr. M. Buratti in performing this work.

They also indebted to the following ISML technicians: ing. Belvedere (plastic transformation); dr. Bracale (preparation of ingots); dr. Farina (stress-corrosion tests); Mr. D. Morri (tensile and fracture toughness tests); ing. Screm (fatigue tests).

This work was sponsored by the Italian Ministry of Defense and the U.S. Department of the Army, under a cooperative research and development project.

REFERENCES

1. W. E. Duckworth: *J. Metals,* 1966, vol. 1, pp. 915-22.
2. V. S. Ivanova: *Met. Sci. Heat Treatment,* 1967, no. 5, pp. 339-45.
3. E. Di Russo: *Alluminio, Nuova Met.,* 1967, vol. 36, pp. 9-15.
4. T. F. Bower, S. N. Singh, and M. C. Flemings: *Met. Trans.,* 1970, vol. 1, pp. 191-97.
5. J. Mercier and R. Chevigny: *Mem. Sci. Rev. Met.,* 1963, vol. 60, pp. 61-71.
6. V. A. Pavlov, Yu. I. Filippov, and S. A. Frizen: *Fiz. Met. Metalloved.,* 1965, vol. 20, pp. 770-74.
7. B. Chalmers: *Prog. in Mater. Sci.,* vol. 10, ed., p. 151, Pergamon Press, Oxford, 1963.
8. J. Nutting: *Met. Trans.,* 1971, vol. 2, pp. 45-51.
9. T. Ikeno: *Nippon Kinzoku Gakkai Sci.,* 1951, vol. 15, pp. 464-66.
10. J. I. Polmear and P. Scott-Young: *J. Inst. Metals,* 1958, vol. 87, pp. 65-70.
11. G. M. K. Sarma and T. R. Anantharaman: *Trans. Indian Inst. Metals,* 1965, vol. 18, pp. 151-54.
12. E. Di Russo: Final Rep. on contract number DA-91-591-EUC 3782, ISML, Novara, July 1966.
13. E. Di Russo: *Alluminio, Nuova Met.,* 1967, vol. 36, pp. 348-63.
14. E. Di Russo: Final Report No. AD-676107 on contract number DAJA37-67-0-0387, ISML, Novara, April 1968.
15. W. Gruhl: *Aluminium,* 1962, vol. 38, pp. 775-79.
16. W. Gruhl: *Z. Metallk.,* 1964, vol. 60, pp. 577-82.
17. D. O. Sprowls: U. S. Patent No. 3,198,676, August 1965.
18. P. L. Mehr, E. H. Spuhler, and L. W. Mayer: Alcoa Alloy 7075T73, Alcoa Green Letter No. 206, Revised August 1965.
19. J. A. Nock: U. S. Patent No. 2,083,576 June 1937.
20. J. C. Bendedyk: *Light Metal Age,* 1968, vol. 26, pp. 10, 12.
21. M. Conserva and E. Di Russo: Report No. 68/18 338, ISML, Novara, August 1968.
22. M. Conserva, E. Di Russo, and A. Giarda: Report No. 68/18 344, ISML, Novara, August 1968.
23. M. Conserva, E. Di Russo, and F. Gatto: *Alluminio, Nuova Met.,* 1968, vol. 37, pp. 441-45.
24. M. Conserva, M. Buratti, and E. Di Russo: Report No. 69/19 247, ISML, Novara, May 1969.
25. E. Di Russo, M. Conserva, and M. Buratti: Report No. 69/19 349, ISML, Novara, May 1969.
26. E. Di Russo and M. Conserva: Report No. 69/19 475, ISML, Novara, July 1969.
27. M. Conserva, M. Buratti, and E. Di Russo: Report No. 69/19 493, ISML, Novara, July 1969.
28. E. Di Russo, M. Conserva, and M. Buratti: Report No. 69/19 496, ISML, Novara, July 1969.
29. E. Di Russo, M. Conserva, and M. Buratti: Report No. 69/19 561, ISML, Novara, July 1969.
30. M. Buratti: Report No. 70/20 616, ISML, Novara, September 1970.
31. I.S.M.L.: Final Report on CTSD-32 Contract, April 1971.
32. E. Di Russo and M. Conserva: Italian Patent No. 886 185, 1971, U. S. Patent Case No. 28514, 1970.
33. M. Buratti, M. Conserva, and E. Di Russo: Report No. 71/21 904 ISML, Novara, November 1971.
34. G. Bracale and A. Alti: Report No. 69/19 563, ISML, Novara, July 1969.
35. G. Bracale: Report No. 71/21 170, ISML, Novara, March 1971.
36. S. N. Singh and M. C. Fleming: *Trans. TMS-AIME,* 1969, vol. 245, pp. 1811-19.
37. H. W. Antes, S. Lipson, and H. Rosenthal: *Trans. TMS-AIME,* 1967, vol. 239, pp. 1634-42.
38. H. W. Antes and H. Markus: *Met. Eng. Quart.,* 1970, vol. 10, pp. 9-11.
39. J. H. Mulherin and H. Rosenthal: *Met. Trans.,* 1971, vol. 2, pp. 427-32.
40. M. Conserva, E. Di Russo, M. Buratti, and F. Gatto: *Mater. Sci. Eng.,* 1973, vol. 11, pp. 103-12.
41. M. Buratti and E. Di Russo: *Alluminio, Nuova Met.,* 1970, vol. 39, pp. 69-75.
42. A. H. Sullivan: *Mater. Res. Stand.,* 1964, vol. 4, pp. 20-24.
43. D. Morri: Report No. 70/20 565, ISML, Novara, 1970.
44. C. Arcolin, M. Buratti, and E. Di Russo: Report No. 71/21 547, ISML, Novara, July 1971.
45. F. Gatto: *Rev. Met.,* 1958, vol. 60, pp. 1085-90.
46. D. Morri: *Alluminio,* 1958, vol. 27, pp. 387-90.