Navy Department - Office of Research and Inventions

NAVAL RESEARCH LABORATORY
Washington, D. C.

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PHYSICAL METALLURGY DIVISION - NONFERROUS SECTION

June 1946

FR-2855

First Partial Report
on
THE DEVELOPMENT OF ALUMINUM
ALLOYS FOR ELEVATED TEMPERATURE
APPLICATIONS

The Effects of 14 Elements on the Short-
time Tensile and Other Properties of 43
Binary and Six Polynary Aluminum Alloys
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Report No. M-2855

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NRL Problem No. M-84
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ABSTRACT

The work contained herein represents completion of the preliminary stage of an overall investigation required in the development of improved aluminum alloys for use at elevated temperatures, especially within a range of 300 to 400°C.

To obtain a better understanding of the relative behavior of various elements in aluminum, the short-time tensile properties at 25, 200, 300 and 400°C. were determined, after stabilizing, for wrought binary alloys of 14 elements. The elements investigated were copper, silicon, nickel, iron, magnesium, manganese, chromium, cobalt, tungsten, molybdenum, vanadium, titanium, boron and cerium (plus other rare earths). Wrought alloys only were covered in this investigation.

The amounts of single additions of each element required, under prescribed test conditions, to give optimum tensile properties at 300 and 400°C. were determined and tabulated in order of decreasing effectiveness (Table 6).

In this investigation magnesium was found to be the most effective strengthener at elevated temperatures. Although at first this element appeared promising as a basis for new alloys, results of an extensive investigation later made available by the Aluminum Company of America showed that wrought polynary alloys, featuring 6 percent magnesium, did not possess better all-round properties for elevated temperature applications than "Y" alloy and were best suited for special purpose rather than general purpose applications.

Next to magnesium, the rare earths gave the most unexpected and promising results. An alloy containing 8.6 percent rare earths (containing 50% Ce) was found to be analogous to the eutectic silicon alloy with respect to structure and formability, but at the same time gave a much higher tensile strength at 400°C. and, in this respect, compared favorably with binary alloys each containing about 5 percent of manganese and chromium. Extrapolated results indicated that the tensile properties at elevated temperatures of the eutectic Al-Rare earth alloy also would be considerably better than those of the eutectic Al-6Ni alloy.

The effects of the remaining elements on short-time tensile properties have been thoroughly discussed with special attention given to manganese, nickel, iron, and cobalt. The order of optimum effectiveness of the elements in improving tensile properties was found to be a function of both composition and temperature. In prescribed amounts nickel and cobalt proved, for example, more effective than manganese at 300°C.; manganese, however, proved best of the three at 400°C.

It was shown that the use of Cr, W, V, and Mo, in more than just grain refining amounts, should not be overlooked in the development of new alloys for use at elevated temperatures, in spite of the foundry difficulties inherent in the use of these elements.

In addition to the investigation of the short-time tensile properties, a detailed discussion is included on the effects of the various elements in aluminum on the microstructures (grain size), formability, fracture(as cast) and hot rupture characteristics of the wrought alloys, and, to a lesser degree, on the microstructure and hardness of the alloys in the cast and wrought condition. Some conclusions in these respects were made.

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AUTHORIZATION

1. Development of aluminum alloys for elevated temperature applications was requested in BuAer ltr. Aer-E-253-NEP to NRL dated 9 January 1946. TED No. NRL-2586 was assigned to this investigation under Project Order No. 578-45. NRL Problem No. M-84 was assigned to the project.

STATEMENT OF PROBLEM

2. The purpose of this investigation was to explore the possibility of obtaining improved aluminum alloys for engine applications in the temperature range of 177° to 400°C (350 to 720°F). The particular applications involved are cylinder heads and pistons in the conventional reciprocating engines, and compressor casings, compressor rotors and compressor blades for jet propulsion units and rockets.

3. To determine the overall suitability of the above materials, the properties which must be considered are thermal conductivity, usual tensile properties, resistance to creep or relaxation, fatigue properties at operating temperatures, corrosion resistance at atmospheric and operating temperatures and dimensional stability. Other properties, such as compressive strength, hardness and resistance to shock and wear may also prove important.

4. In light of recent developments in aeronautics the problem of developing aluminum alloys having better properties in the temperature range specified is becoming increasingly important. Designers are placing more and more emphasis on greater power and greater operating efficiency in power units to compensate for the increased weight and/or speed now in demand. This, of course, means higher operating temperatures, and it is not a remote possibility that the day will come when the maximum power and speed obtainable in lighter-than-air-craft will be governed to a great extent by the strength of the materials in and near the power units. In order to perform a given job satisfactorily at elevated temperatures, an alloy, to be suitable, must possess an aggregate of properties, such as good thermal conductivity, strength and resistance to creep and others, each of which must meet a prescribed minimum. Herein lies the crux of the problem facing investigators in this field. During the past decade alloys have been developed having one or perhaps two exceptionally good properties at elevated temperature, but only at a disqualifying sacrifice of other properties of equal importance.

SCOPE OF INVESTIGATION

5. This investigation was a preliminary part of an overall investigation required to develop an alloy or alloys having the necessary

properties, as discussed above, to perform the necessary functions satisfactorily at elevated temperatures, especially within the range of 300 to 400°C. In view of the fact that little quantitative information can be found in the literature* pertaining to the effects at elevated temperatures (above 200°C) of various elements in aluminum and, since, in an investigation of this type it was considered necessary to make a complete coverage of all feasible elements, it was believed the logical approach to an overall investigation would be to determine the effects of single additions of various elements to aluminum with the purpose of obtaining quantitative, semi-quantitative and qualitative information which would offer a reasonably concrete basis for selecting compositions for future investigations of polynary alloys. With this plan in mind fourteen elements were selected. A few polynary alloys were made during the latter part of the investigation and these results also have been included.

6. The elements whose effects on aluminum were investigated were copper, silicon, nickel, iron, magnesium, manganese, chromium, cobalt, tungsten, molybdenum, vanadium, titanium, boron and cerium, plus other rare earths. It was decided not to investigate binary alloys with beryllium, as a survey of the work on aluminum beryllium alloys, compiled by Jeffries (2), revealed that these alloys have not come up to expectations. That straight aluminum-beryllium alloys offer no advantage over numerous commercial alloys now used at elevated temperatures, had been substantiated also by the Bureau of Standards (3). A few polynary systems containing beryllium have proved promising, however, and should be investigated further.

7. The properties measured or observed in this investigation are as follows:

1. Ultimate tensile stress at room and elevated temperatures.
2. Tensile yield stress (0.2% offset) at room and elevated temperatures.
3. Elongation in 2 in. at room and elevated temperatures.
4. Surface quality and appearance and shrinkage characteristics of permanent mold castings.
5. Fracture characteristics of as cast metal.
6. Macrostructure (grain size) of as cast metal.
7. Microstructures of as cast alloys.
8. Macrostructure and microstructure of selected as rolled alloys before and after stabilizing and testing at elevated temperatures.
9. Hot workability.
10. Characteristics of metal ruptured at room and elevated temperatures (strip tensile specimens).

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*The "Alcoa" (Aluminum Company of America) report on this subject (1) was not released until after the final stages of this preliminary investigation were nearly completed. Practically all previous information came from the British. For comparison, frequent mention of the Alcoa results will be made throughout this report.

11. Hardness of as cast alloys and of as rolled alloys - the latter before and after stabilizing and testing at elevated temperatures.

All tensile tests were made with a free running crosshead speed of 0.05 in. per minute up to and just beyond the yield load and from 0.1 to 0.15 in. per minute from yield load to ultimate load and at a maximum of 0.3 in. per minute from the ultimate load to rupture. Temperatures investigated included room temperature (ave. 25°C) 200, 300 and 400°C., after a sufficiently long stabilizing treatment at 210, 210 and 410°C., respectively, for periods ranging from 5 to 24 days. Each alloy was tested at room temperature in the stress relieved condition and, in some cases, additional specimens in the as machined condition, or estimated 1/2 H temper. Although in an overall investigation on this type of problem both as cast and wrought alloys should be considered, lack of time permitted wrought alloys only to be included in this report. Likewise comparisons were to be made under identical test conditions with several commercial alloys now used for elevated temperature applications, but, unfortunately, these investigations were not completed.

KNOWN FACTS BEARING ON PROBLEM

8. As a result of the remarkable progress in light alloy technology, especially within the past fifteen years, there is now abundant literature concerning the effects of various addition elements, singly and collectively, on aluminum at room and moderate temperatures (below 200°C). Comparatively little information, however, has been published on the effects of addition elements at elevated temperatures (300° to 400°C). The most noteworthy work in this field was carried out by the British, namely Brimelow (4, 5, 6, 7), and later by Kempf, et al (1) of Alcoa, whose work was just recently made available. Both of these extensive investigations on the development of new and better aluminum alloys for elevated temperature applications were started about 1930. Sufficient information has come from Germany through various sources to indicate that German scientists have long been thinking and experimenting along these lines and, out of the necessity of having to develop substitute materials, ferrous and nonferrous alloys with unusual and in some cases seemingly unorthodox compositions have been reported, both in German technical literature and in reports on the examination of captured enemy equipment. Of the newer German aluminum alloys used for elevated temperature applications, however, none has been noted, except for one which will be discussed later, which has compositions or properties varying appreciably from those of the conventional alloys developed since World War I.

9. A survey of the properties of standard alloys commonly used in this country for elevated temperature applications was made by Wyman (8) in 1943, and since the present investigation is concerned

primarily with new types of alloys, further discussion of conventional alloys is unnecessary. British and German commercial alloys for elevated temperature use, in general, have properties somewhat similar to the alloys used in this country even though there may be minor variations in composition. A brief advisory report prepared by Jeffries (9) contributed much to a better understanding of the effects of various elements (impurities) on the corrosion resistance of aluminum alloys. Although this report dealt with standard type alloys for use at low temperatures, most of the information would also serve to indicate behavior at elevated temperatures.

10. New experimental alloys for elevated temperature applications to be given production and operating tests were first reported on by British investigators (4, 5, 6, 7). This work was done at the Royal Aircraft Establishment in cooperation with the National Physical Laboratory and other organizations. Five alloys were selected from the overall preliminary investigations for further tests and were designated as 40B, 47D, 47B, 40C and 55. Of these the latter three looked most promising. Their nominal compositions are as follows:

		Per cent									
Alloy	Type	Cu	Ni	Mn	Mg	Fe	Si	Cr	Ti	Be	Bal.Al.
47B	Cast	1	4	3	0.5	0.5	0.2	-	-	-	"
55	Cast or Wrought	1.8	2.7	2	0.6	0.4	0.2	0.3	0.005	-	"
40C	Wrought	2	5	2.5	0.6	0.5	0.2	0.5	-	0.4	"

The new alloys in the cast condition were found to possess at 350°C and higher temperatures higher Brinell hardness, tensile strength and resistance to creep and wear than existing British standard aluminum alloys. The new alloys have rather low thermal conductivities, especially in the cast condition and slightly greater densities than the alloys normally used for pistons and cylinder heads. Their machinability was considered to be excellent. In the wrought condition the new alloys had slightly lower tensile and fatigue strength compared with the standard alloys. The behavior of wrought and chill cast pistons of the selected alloys was investigated. The experimental pistons gave a favorable performance in comparison with standard pistons tested under similar conditions; the wear-resisting properties of the new alloys appeared to be of particular value in avoiding excessive ring-groove wear. A more detailed abstract of this work, including graphs and tables, may be found in Wyman's report (8).

11. The German alloys referred to in the latter part of Paragraph 8 were developed by Rontgen and Koch (10) and have the following nominal composition:

Per cent						
Cu	Co	Cr	Mo	Ni	Others	Al.
4	1	1	1	1	0.5	Bal.

This alloy was stated to have a tensile strength of 12000 psi. at 400°C. The alloys must have had some unsatisfactory characteristics, however, as no reference to its being used commercially for elevated temperature applications has been noted.

12. While working on new aluminum alloys for use at elevated temperatures, Bollenrath and Grober (11) investigated a number of alloys containing appreciable amounts of magnesium, two of which were used mainly for cylinder heads in German aero-engines during World War II. These alloys are known as Hy 51 and Hy 511 and are generally used in the wrought or sand cast condition. Their nominal compositions are as follows:

	Per cent						
Alloy	**Cu**	**Fe**	**Mn**	**Mg**	**Si**	**Ti**	**Zn**
Hy 51	0.08	0.4	0.25	5	1	0.01	.0.06
Hy 511	0.02	0.35	0.19	4.92	1.16	0.13	0.06

An idea of the tensile properties at <u>room temperature</u> of this type of alloy after <u>stabilization</u> at various elevated temperatures can be found in the following data on the forged Hy 51 alloy:

	Temperature of Stabilization							
	270°C				320°C			
Stabilized at Temp.	Y. S. 0.2% offset psi	T. S. psi	% elong. in 10 x dia.	BHN 5/250/30	Y. S.	T. S.	Elong.	BHN
40 hrs.	16,050	34,580	16	63	13,580	33,050	19.8	64
340 hrs.	11,620	32,620	20	65	13,160	33,050	18.4	64

All tests at room temperature

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13. The results of the extensive Alcoa investigations (1) on alloys for elevated temperature use, featuring relatively high amounts of magnesium and other elements as compared with standard alloys, indicated three types of alloys worth further consideration. The nominal compositions of these types are as follows:

	Per cent						
Type	Mg	Mn	Ni	Cu	Si	Fe	Al
(1)	6	1	1.5	0.4	0.5	0.5	Bal. (254 wrought alloy)
(2)	10	-	-	8	0.5	0.5	Bal. (cast alloy)
(3)	-	-	4	12	7	0.5	Bal. (modified 177 cast alloy)

The third alloy represents the best modification of the alloy formerly designated as 177 - a Al-7Cu-7Si-7Ni-type alloy. Alloy 254 was once put on the market but was later recalled as there was no demand for it. It is not claimed that these alloys are better than the standard alloys now used when the overall properties necessary for satisfactory performance at elevated temperature applications are taken into consideration. It is stated, however, that the 254 alloy has been developed to the stage where it may be used, with or without substitution of 1.5% cobalt for nickel, for specific applications wherein a sacrifice in certain properties may be tolerated to obtain a maximum in a specific property.

14. The aluminum-silicon eutectiferous type alloys, because of their potentially desirable structure for elevated temperature applications, were of the first to be extensively investigated. The work of Dudzinski and Brimelow (12, 13) is one of the most recent and complete investigations in this field and deals mainly with alloys containing 10-14% Si, 1.5-6% Cu, 0.3-0.7% Mg, 0-2% Ni, 0.2-0.3 Co, 0.05 - 0.15 Ti, 0.4 - 0.5 Fe and 0 - 0.2 Mn. In this paper the work of several other investigators, dealing essentially with various ramifications of and minor additions to the foregoing composition range was reviewed; namely, that of Kotzschke, C Hisatsume, Itami, Bauer and Winterhager and Bollenrath. The results indicated that up to the present time alloys of this type are expected to perform satisfactorily only to moderate temperature (250-280°C) and, in general, for overall performance do not surpass the commercial A132 type alloy, containing 12 Si, 0.8 Cu, 1 Mg and 2.5 Ni and other standard alloys.

THEORETICAL CONSIDERATIONS

15. Any project intended to investigate thoroughly the possibility of developing improved aluminum alloys for use at elevated temperatures would necessarily be undertaken from the following approaches:

a. Effect of alloying elements in solid solution.
b. Effect of alloying elements with no or limited solid solubility.
c. Effect, in precipitation hardening alloys, of employing elements of low or negligible diffusibility to correlate with expected life of parts.
d. Eutectic type alloys having a strengthened and stable matrix.

The trend of recent investigations is mainly toward approach (c). Research carried on thus far seemed to indicate that the type of structure most suitable for strength and hardness at elevated temperatures was a network of fine, hard particles evenly dispersed in a matrix (solid solution or Al-rich phase) having a melting point of about 600°C or above (4). It has long been known that the properties of "Y" alloy, in common with other aluminum alloys, depend chiefly on the type and state of aggregation of the heterogeneous phases. Some commercial aluminum alloys have as high as 20 per cent of their volume made up of heterogeneous inclusions. The alloy content of some of the new experimental alloys indicate that this percentage would be increased considerably. The distribution and character of the inclusions is governed mainly by the composition, method of casting, amount of working and heat treatment. To a large extent the character of the heterogeneous phases, i.e., whether angular, fibrous or rounded, determines the resistance of the aluminum alloys to softening, wear and creep at high temperatures. The above factors also influence the forgeability and cold working properties.

16. On the other hand an aggregate of additional phases is, in most cases, accompanied by an increase in density, a decrease in thermal conductivity and a tendency toward undesirable hot and cold working qualities and foundry characteristics (there are exceptions). Structural and, hence, dimensional stability, fatigue properties and resistance to corrosion may or may not be enhanced by the presence of heterogeneous phases. The net result is that an alloy, to perform a given job satisfactorily at elevated temperatures, must represent a compromise of the foregoing properties and characteristics.

17. Other hypotheses are thought provoking in regard to the elastic and plastic behavior of metals at elevated temperatures. One is the question of grain size. It is generally accepted that in most cases, depending on type of alloy, weakening of polycrystalline alloys under stress at elevated temperatures occurs first in the grain boundry material (phase). If such is the case, one would expect a large grained material, because of the smaller surface area, to offer better resistance to creep than a fine grained alloy. This has been reported to be the case for some alloys. A coarse grained alloy, however, is conducive to poor workability, which would make it undesirable in a good many cases to start a series of forming operations with a large grained casting. Why then, would it not be feasible to increase the hot strength by recrystallizing the finished product? Recrystallized structures, heat treated to optimum grain size, are generally coarser grained than those in the as worked or as cast state. That recystallized structures offer greater resistance to creep, however, is not in accordance with general opinion and experimental evidence is lacking to definitely clarify the situation. The effects of degree of deformation, temperature and initital grain size on rates of nucleation and growth have been investigated considerably and results are in common agreement; however, much remains to be determined on the effects of composition on these phenomena. To obtain stable alloys for elevated temperature use, the trend of thought is toward the addition of elements which tend to retard the recrystallization rate (by hindering nucleation) and to increase the recrystallization temperature. This implies that to be suitable, the structure obtained thereby should be stronger at elevated temperatures than a larger grained, recrystallized structure. It is known that recrystallization is markedly affected by elements

entering into solid solution. In general, such materials tend to recrystallize at higher temperatures and it is very probable that either the rate of nucleation or rate of growth, or both, are affected by presence of additions forming solid solutions (14).

18. The mechanism and amount of hardening of aluminum alloys undoubtedly plays an important part in governing mechanical properties at elevated temperatures. Although the exact rate still remains obscure, it would fall into two general categories; namely, the effects due to elements in solid solution and those due to compounds or heterogeneous phases formed by elements with no or limited solubility. It is not known whether the latter phases, in spite of their size and character, would have effects entirely independent from those of the solid solution. From the study of macro and microstructures the effects would seem to be both independent and mechanical, except for their decreasing or otherwise influencing the solid solubility, through selective compound formation, of other elements.

19. Many theories have been advanced concerning the behavior of grain boundry material under stress at elevated temperatures. One is that the weakening of the grain boundry material may be attributed mainly to (1) its lower melting point, and (2) subsequent softening, which is caused by a weakening or breakdown of the lattice bonds through the increase in internal energy. The amount of softening would vary with time as the composition of the boundry material changed (and the equilibrium of forces within the lattice upset) through the process of diffusion of atoms into the surrounding grains. The rate of diffusion may be influenced by several factors, such as temperature, deformation, chemical dissimilarity of atoms and degree of atomic disorder within the lattice. On the other hand, advocates (if any remain) of the amorphous cement theory, in which the boundry phase is considered to behave plastically, minimize the possible effects of diffusion on the creep characteristics of a stressed polycrystalline alloy, principally on the grounds that any weakening and subsequent creep due to diffusion would be negligible as compared with that resulting from the reorientation of grains and slippage on planes within grains surrounded by a boundry material of plastic behavior. Other criticisms of the diffusion effects have been extremely fundamental in nature and deal mainly with the factors affecting the activation energies which govern the "eligibility" of atoms for and probability of self-diffusion. Such factors involved are the location of the atoms with reference to lattice imperfections such as grain boundaries and other atoms and molecules and temperature (15).

20. The diffusion concepts under various conditions of stress and temperature have been treated mathetmatically by several investigators and, in general, are relatively involved and beyond the scope of this brief survey. A detailed report on the diffusivity of elements in aluminum, the various methods used in their determination and the influence of diffusion on the properties of metals at elevated

temperatures, based on concepts and data available to date, has been prepared at this laboratory by Ciborski (16).

EXPERIMENTAL PROCEDURES

Preliminary Considerations

a. Screening-out test

21. At the beginning of this investigation it was realized that a large number of binary alloys would have to be made in order to determine the effects of all feasible addition elements to aluminum. Consequently, considerable thought was given to the type of screening out test to be used in evaluating these alloys. Brimelow (4) utilized Brinell hardness tests at room and elevated temperatures, together with the chemical compositions and cooling curves of the alloys as "a criterion of mechanical strength for the purpose of making a rapid survey of the alloys". Kempf, et al, (1) also used the hot hardness test, as well as fracture tests, to make a rapid survey of binary alloys. From the standpoint of type of specimen required, the ease of conducting tests and the time involved, the foregoing tests would be ideal for screening-out purposes provided they gave pertinent information. Analysis of the data contained in (4), however, failed to show any definite trend of behavior in hardness, with the exception of a decrease in hot hardness to be expected with an increase in testing temperature. That is, a number of alloys which were not in the group selected for further investigation had hot hardness properties equal to some alloys that were. Further analysis of these and other data also failed to show a definite relationship between tensile strength and hardness at elevated temperatures. It was realized that for some alloys, especially those with elements in solid solution, the hot hardness test might indicate their potentialities of having better strength at elevated temperatures than others. For wide coverage of alloys, however, the hot hardness test was not considered suitable for the intended purpose.

22. After further consideration it was decided that the most feasible screening-out test would be a short-time tensile test at various elevated temperatures after a sufficiently long stabilizing treatment. Considerable time would be involved, but it would be relatively short compared with that required for stress rupture, creep and other tests that subsequently must be made on selected polynary alloys. Although the short-time tensile test offers a reasonably reliable criterion fc evaluating and comparing the various alloys, it is to be realized tha data obtained therefrom have their limitations, especially with respect to their engineering significance to the designer.

b. Type of Test Specimen

23. With specimens of each alloy to be tested at least in duplicate

at each temperature, it was apparent that a large number would be required and, hence, considerable thought was given to the type of specimen used. From the standpoint of material consumption, ease of machining and overall expense, it was decided to use sheet specimens 8 inches in length, having 2 in. gage lengths, conforming to the A.S.T.M. Standard B25-38T (Ref.20), as shown on Plate 1. The advantages in using this type of specimen, in contrast to the round type, were considered to outweigh the disadvantages; namely, the increased amount of preliminary fabrication required and the greater difficulty encountered in testing strip specimens at elevated temperature. Special loading bolts and techniques were required to set up specimens of insufficient length to extend through the furnace.

c. Selection of Compositions

24. The compositions investigated were selected to conform with the scheme presented in Paragraph 15 as far as was consistent for the particular alloy system concerned, that is to say, on the basis of equilibrium diagrams and existing information. It would not be feasible, for example, to investigate binary eutectic systems of Al-Mg or Al-Cu, containing 33 per cent of Mg and Cu, respectively, for their properties are known to be disqualifying for the intended purpose. The properties at room and moderate temperatures of a number of the alloys made were already known but not at temperatures ranging from 300 to 400°C.

25. The general plan followed in selecting compositions of binary alloys, both for elements having solid solubility and limited or no solubility, was to use amounts that would give a representative picture of the effects of a given element at low, moderate and high percentages - the latter usually considerably higher than reported in standard alloys. Eutectic alloys were included when feasible and special interest was taken, in the case of those elements having appreciable or limited solid solubility, to the relative amount that would be in or out of solution at 400°C. In some cases only two or three alloys, covering a wide composition range, were made to investigate the effects of a particular element and as such the strength-composition curves that were plotted may indicate no more than a trend, although in general no erratic deviation should be expected from the curves as plotted.

Materials Used

26. The fifteen elements used in the preparation of heats were of the highest purity obtainable. Master alloys of aluminum were used for all additions except magnesium. The compositions of the master alloys are given in the latter part of Table 1. The relatively high percentages of iron reported for the alloys of titanium, tungsten, molybdenum and beryllium may be attributed mainly to pickup from the stirring rod at the high temperatures required; all master alloys were made at this laboratory except those of Al-V and Al-Ce, which were obtained from Alcoa and the Cerium Corporation of America, respectively.

Melting and Casting Procedure

27. The alloys were melted in 6 lb. heats (unless otherwise noted) in a high frequency induction furnace. Number 17 clay crucibles were used. About three-fourths of the charge of high purity aluminum (99.95%) was melted first, followed by the master alloys in order of decreasing melting point (for polynary alloys). Magnesium and the lower melting master alloys were added at the lowest possible temperature, usually after the melts had been cooled by adding the remaining aluminum. Some heats were presolidified for gas elimination when high operating temperatures were necessary; others were later remelted because of dirty metal or other undesirable characteristics indicated by fracture test of the as cast metal. All heats were degassed four minutes with dry nitrogen before pouring; no flux was used. The temperatures were observed by means of a chromel-alumel direct immersion thermocouple with Leeds and Northrup pyrometer having an accuracy of ± 5°C. The metal was cast into a permanent mold in slabs 8 inches in length, 1-1/4 inches in width and 5-3/4 inches in height plus a 1-3/4 inch half-flare at the top to aid feeding. The wall thickness of the cast iron mold was 1 inch. The mold was tilted 30 degrees during the pouring (direct from the furnace) to minimize turbulence and returned immediately to the horizontal position. The temperatures of the mold before pouring and of some castings before quenching were determined by a Leeds and Northrup contact pyrometer. Complete details of the melting practice for each heat are given in Table 2.

Preparation of Specimens

28. Billets 4-1/2 inches in length x 1-3/4 inches in width and 1-1/4 inches thick were cut from the bottom of each casting for hot working by forging and rolling into strips for the tensile specimens. The general procedure used in working down the cast alloys was as follows:

1. The billets were held approximately 16 hours at 438°C. before forging to a thickness of 1/2 inch, the forging temperature range being about 430 to 380°C. Intermediate anneals of 10 to 20 minutes duration were given, as required, to some of the harder alloys.
2. The forged slabs, now approximately 14 in. x 2-1/4 in. x 1/2 in. were soaked for a minimum of 2 hrs. at 390 to 416°C. before rolling to an average thickness of 0.095 in. The rolling temperature range was about 400 to 360°C. Intermediate anneals were given when necessary.
3. A few alloys, especially those containing high percentages of Mn, Cr, and Cu, were forged to a 10 to 15% reduction of area and then homogenized at 440 to 480°C. for 16 to 24 hrs. before forging to 1/2 in.; additional anneals were necessary.

Specimen blanks 8 in. x 3/4 in. were cut from each rolled strip and stress relieved for about 2 hrs. at 315°C, (600°F) before final machining.

Heat Treatment and Stabilization of Specimens

29. All specimens tested at room temperature, except those containing copper, were in the stress relieved condition (315°C, 2 hrs., F.C.). The copper bearing alloys were tested in the solution treated and **annealed** condition. In addition, the alloys containing magnesium, nickel and iron were tested at room temperature in the as machined condition and, therefore, in an estimated 1/2 H temper.

30. Specimens tested at 200, 300 and 400°C, were stabilized at 210, 310 and 410°C., respectively, for periods ranging from 5 to 24 days, depending on the temperature. As before, specimens of alloys containing copper, manganese and chromium were solution treated before stabilization. Unless otherwise noted, specimens were cold water quenched from the stabilization temperature. An account of the specific heat treatment or treatments given each alloy is given in Table 3 along with the tensile properties.

Tensile Testing

a. Load and Extension Measurement

31. Tensile testing was done with a hydraulic Rhiele tensile machine of a 30,000 lbs. capacity having a mechanical type stress-strain recorder. For tests at elevated temperature, a furnace was mounted between the crossheads, with the loading bolts extending from each crosshead into the furnace; whereas, in the tests at room temperature, the furnace was removed and the strip specimens inserted in grips contained in the crossheads. This was done to avoid excessive strain on the loading bolts due to the higher loads at room temperature. Except when testing a few of the strongest alloys at room temperature, the lowest (3000 lb.) scale on the machine was used, and the maximum error in load readings as a result of minor variations in initial settings and crosshead speed, was thought to be no more than 5 to 17 lbs.

32. The load was transmitted directly to the stress-strain recorder by means of an arm from the pendulum or lever arm of the tensile machine, whereas the strain was transmitted to the recorder at a 16 to 1 magnification by means of a 28 gage copper wire fastened to the free-running crosshead. This type of extensometer is not conducive to high accuracy in obtaining yield loads as it would include any strain that might occur outside the 2 inch gage length of the specimen as well as any fluctuation in the free-running crosshead. The latter, in fact, was found to have some effect on the stress-strain curve.

In order to investigate the accuracy of the Rhiele machine, several alloys exhibiting questionable mechanical properties were tested at room temperature on the larger tensile machine with a more accurate electrical extensometer connected to the gage length. Results indicated that the magnitude of error in the yield strengths reported should not exceed ± 5 per cent. Due to the lower yield loads, the questionable results on the stress-strain curves were not obtained for the alloys tested at 300 and 400°C. Some of the curves were, nevertheless, questionable and these yields strengths were reported as such in the table of tensile data. Yield strengths were determined by means of the 0.2 per cent offset method.

b. Heating and Temperature Control

33. Following stabilization at 10°C above the test temperature and usually short periods of storage, the specimens were tested at elevated temperature in an automatically controlled 13 in. x 7 in. diameter resistance furnace having a 1-3/8 in. diameter center tube. The furnace could be opened (book-type) to facilitate the handling of specimens. Specimens were brought to temperature in approximately 15 to 20 minutes and were held at temperature for a minimum of 20 minutes, before the test was started. The holding time was sufficiently long to obtain thermal equilibrium in the stabilized strip specimens. All tensile tests were made with a free running crosshead speed of 0.05 in. per min. up to and just beyond the yield load and at 0.0 in. per min. (unless otherwise noted) from the yield load to the ultimate load, and a maximum of 0.3 in. per min. from the ultimate load to rupture. In no case was the crosshead speed increased above 0.1 or 0.15 in. per min. until the load had started to drop off. The speed was increased primarily to decrease the testing time and, hence, the possibility of temperature fluctuations.

34. Since the furnace contained only one bank of heating element and no external heating appliances, extremely accurate thermal control was not obtainable. A survey, not made during a tensile test, of the distribution of temperature over the central 3 in. of a specimen, however, showed a temperature variation no greater than ± 30°C. at 400°C. (usually less), which complies with A.S.T.M. Standard E21-37 (21). The temperature measurements for the survey were obtained with shielded 28-gage chromel-alumel thermocouples flash welded to the specimen. Later temperature measurements made during the tensile

tests indicated that with those specimens of high elongation, especially where the test time was at least five minutes, a decrease in temperature of 5 to 10°C. often occurred in the upper part of the specimen due to the decrease in temperature of the upper loading bolt, more of which was continually leaving the environs of the furnace. This decrease, however, always occurred considerably after the yield and ultimate loads had been reached and, therefore, would be reflected only in the overall elongation. No further investigation was made to determine the actual effect, if any, on elongation. For test runs, two 28 gage iron-constantan thermocouples were flash welded to the grip ends of the specimen, each about 1/4 to 1/2 in. from the reduced section. A double throw switch was fastened to the potentiometer so that both millivolt readings could be taken in a minimum of time; several readings were taken during each test.

c. Loading Bolts

35. The design of an entirely satifactory loading bolt for testing strip specimens was hampered by the maximum allowable diameter. The screw-type friction grips, as shown in Fig. 2, Plate 2, were first used and although these grips were satisfactory for tests at 300 and 400°C, some slippage occurred in testing the stronger alloys at 200°C. Such slippage was conducive to strain hardening during the course of the test. To eliminate this difficulty new self-clamping, wedge-type friction grips were designed, as shown in Fig. 3, Plate 2. These grips proved satisfactory for tests at 200, 300 and 400°C.

Hardness Tests

36. The Brinell hardness was determined for selected as-cast alloys, using a 10 mm ball and 500 kilogram load for 30 seconds. The Brinell test could not be used for the wrought alloys as the strip specimens were too thin. Rockwell hardness measurements were made on the strip specimens in the stress relieved or heat treated condition (as tested at room temperature) and after stabilizing and testing at elevated temperature, using the RH scale for all alloys and the RE scale also for the harder materials. Because of the wide range of hardness values encountered with the binary alloys, the RH was the only scale that enabled a complete coverage of all the alloys, and although some of the values obtained were too high or low on the scale to be truly significant, they nevertheless served to give a trend of the relative hardness of the binary alloys.

Other Tests and Observations

37. In addition to the measurement of the properties previously discussed, other modes of behavior and characteristics of all the alloys were noted throughout the investigation. Fracture tests were made with as-cast metal to obtain an indication of the elongation, toughness, resistance to shock and soundness of metals. Representative

cross-sectional surfaces (1-3/4 in. x 1-3/4 in.) of cast metal were macro-etched for grain size; microstructures of selected alloys in the as-cast condition were also studied, as well as macrostructures and microstructures of selected alloys in the as-rolled condition - the latter both before and after stabilization and testing at elevated temperatures. In the metallographic work special attention was given to the effects of hot working the as-cast alloys with respect to the size, form and redistribution of constituents and also to the effects on structure of sustained heating at elevated temperatures on the constituents. The surface quality and appearance and shrinkage characteristics of the permanent mold castings were noted and also the hot formability by forging and rolling. Although the tests and observations just numerated are mainly qualitative or semi-quantitative in nature, it was believed that they would contribute materially to a better understanding of the effects of the various elements on aluminum.

RESULTS

38. The results obtained quantitatively in this investigation, namely, chemical compositions, melting practice, tensile properties and hardness, are listed in Tables 1, 2, 3 and 5, respectively. Some of the more pertinent interpretations of the data on selected alloys can be found in Tables 4 and 6 and Plates 3, 4 and 5. Information of a qualitative nature, i. e., characteristics of fractures of as-cast and hot-ruptured specimens, formability and grain size also has been included.

DISCUSSION OF RESULTS

Introduction

39. Since this investigation deals primarily with tensile properties at elevated temperatures, these data will be discussed first. To simplify the discussion, the alloys have been divided into three groups:- the first group includes the aluminum base binary alloys of copper, magnesium, silicon, iron, nickel and manganese; the second includes the binary aluminum alloys of chromium, cobalt, tungsten, molybdenum, vanadium, titanium, boron and cerium and the third deals with the few polynary alloys investigated. Special emphasis will be placed on the behavior of the alloys at 400°C. Curves showing the relationship between the tensile properties at each testing temperature and the per cent additions of each element were plotted from the data in Table 3. Since these curves merely represent a repetition of data they were not included in the report. They did, however, offer a more ready interpretation of the tensile data and were especially useful in determining where the degree of effectiveness of each addition element changed. For these reasons the curves have been mentioned frequently in the discussion and are generally referred to as the "strength-composition" curves. Except for a few

cases, the effects of the addition elements have been discussed in the same order as they are listed in Table 3.

40. As was pointed out in Paragraph 25, only a few alloys were made in some cases to investigate the effects of a particular element and as such the strength-composition curves may indicate no more than a trend. Although at least two specimens per alloy were tested at each temperature, it is realized that additional specimens from several castings of similar composition, made under identical conditions, should be tested before specific compositions or composition ranges are reported as giving certain properties. The discussion in this paper, however, has been based on the actual data reported, as obtained under the prescribed test conditions, and the strength - composition curves plotted therefrom. The mechanical properties given in comparing the effects of the various addition elements should be taken in light of the above considerations.

Additions of Cu

41. From analysis of the data in Table 3 it is seen that for the Al-Cu alloys (1-4) the maximum effectiveness on tensile and yield strengths at 400°C. was obtained at about five per cent copper, whereas, at 300°C. the addition of 3.3 per cent copper gave a tensile strength nearly equal to and a yield strength slightly better than the alloy containing 8.7 per cent copper. The lower yield strength of the high copper alloy may most likely be attributed to the greater amount and larger size of the excess precipitate. At 400°C. no appreciable change in elongation was obtained with additions of copper over 3 percent. As would be expected the highest room temperature properties reported in this investigation were obtained with the artificially aged Al-Cu alloys. It was also interesting to note the rapid decrease in tensile properties at 200°C. with additions of copper exceeding five to six per cent - a phenomena undoubtedly related to the effects of over-aging as well as to the comparatively high amount of precipitate initially present in the Al-8.7 Cu alloy.

42. Single additions of copper, ranging from 2.8 to 8.7 per cent, proved to be surprisingly ineffective as strengtheners at 300 and 400°C., especially at the latter temperature. Comparatively high elongations, however, were obtained with the copper additions at all temperatures. It was evident, that, in polynary alloys containing copper as the principal addition, at least 50 to 60 per cent of the strengthening could be attributed - in contrast to the Al-Mg type alloys - to the effects of other elements. That strength is only one of many qualities required for general purpose applications at elevated temperatures is only too well shown, however, by comparing "Y" alloy with Alcoa 254 alloy.

Additions of Si

43. At both 300 and 400°C. little variation was found in the tensile properties of the alloys containing 0.25 to 0.4 per cent silicon

(alloys 5, 6, 7). At 300°C. the tensile and yield strengths of these alloys leveled off, averaging 2000 and 950 psi, respectively, whereas the corresponding values for the nearly eutectic silicon alloy (No. 8) were 110 and 190 per cent greater. At 400°C. their tensile and yield strengths averaged 930 and 500 psi, respectively, the former being only slightly higher than the tensile strength of pure aluminum and the latter the same as that of aluminum. The corresponding properties at 400°C. for eutectiferous alloy No. 8 were 90 and 140 per cent greater than for the alloys containing up to 1.4 per cent Si. It was evident - and it is well known - that for Al-Si type alloys at least 10 per cent silicon would be necessary for maximum effectiveness for elevated temperature applications. The effects of small amounts of silicon was relatively pronounced, however, at room and moderate temperatures (below 200°C).

Additions of Ni, Fe and Mn

44. A study of the strength-composition curves plotted with data in Table 3 for alloys containing additions of nickel, iron and manganese did much to reveal the relative behavior of these alloys at moderate and higher temperatures. The alloys containing these elements continued to increase in tensile and yield strength with an increase of each element within the composition and temperature ranges tested. The rate magnitude of increase at a given temperature, however, varied considerably with each element. At room temperature, with alloys in a stress relieved condition, the superiority of nickel was clearly indicated; the tensile strength for alloys containing nickel and manganese increased in a nearly linear manner with additions over one per cent, the rate of increase with the additions of nickel being approximately 100 per cent greater than with manganese. The rate of increase in tensile strength at room temperature with increases in iron content up to about 2 per cent was also linear and even greater than that of nickel; beyond this amount the rate dropped off rapidly. At room temperature all of these alloys were characterized by unusually low yield strengths, part of which may be attributed to the particular crosshead speed (0.05 in./min.) used up to the yield point. Alloys containing nickel and manganese held up to the yield point. Alloys containing nickel and manganese held their load well during the course of the tensile test, both at room and elevated temperatures, and were sensitive to changes in crosshead speed. The effects of strain hardening on room temperature properties of alloys containing nickel, iron and magnesium is also indicated in Table 3.

45. The relative degree of effectiveness of additions of nickel, iron and manganese varied considerably at 300 and 400°C., as compared with the effects at 200°C. and below. At 300°C. the strength-composition curves indicated that with additions of the order of 4 to 6 per cent nickel gave tensile strengths from 4 to 24 per cent, respectively, higher than equivalent amounts of manganese, whereas with additions of about 1 to 3 per cent, manganese gave from 18 to 10 per cent,

respectively, higher tensile strengths than equal amounts of nickel. Variations in yield strength were of the same order of magnitude. The curves also showed that the tensile and yield strengths obtained for binary alloys containing about 3-1/2 per cent manganese and 3-1/2 per cent nickel were nearly equal at 300°C., although the former showed an 18 per cent lower elongation. At the same temperature, the tensile strength obtained with an addition of 2 per cent iron was about 95 per cent of that obtained with the 5 per cent addition. At 400°C. manganese proved more effective than nickel in increasing the tensile and yield strengths for all compositions tested. The relationship between strength and composition was nearly linear in each case. As judged from the strength composition curves, additions to aluminum of 2, 4 and 6 per cent of manganese proved about 33, 28 and 20 per cent, respectively, more effective than equivalent amounts of nickel in increasing the tensile strength at 400°C.; however, the tensile-yield ratios for the alloys containing nickel and manganese were about equal. It would be difficult to say just what percentages of manganese and nickel would prove most effective in increasing tensile strength of binary alloys at 400°C., for, within the composition range tested, their strength increased with an increase in addition element. Other properties such as formability, thermal conductivity and hardness would be the governing factors. Alloy No. 11, having the eutectic composition of 5.88 per cent nickel, could be formed with considerably less difficulty than alloy No. 19, containing 5.3 per cent manganese and, therefore, from the standpoint of formability, additions of nickel up to 6 or 8 per cent would be feasible whereas manganese would be limited to maximum additions in the order of 5 to 6 per cent for wrought alloys. The most practical additions for each element, based on overall effects, would be 3.5 to 5 per cent. At 400°C. additions of iron over 1.5 to 1.8 per cent would not be warranted as far as improving the tensile strength is concerned. It was interesting to note that little change in elongation occurred at 300 and 400°C. for the alloys containing nickel and iron.

Additions of Mg

46. The last element of the first group to be discussed is magnesium. From the data in Table 3 it is seen that for the binary alloys the highest tensile and yield strengths at 300 and 400°C. reported in this investigation were obtained with alloy No. 17, containing 5 per cent magnesium. Although the strength-composition curves for magnesium were plotted through only three points, there was a definite indication that the relationship at 400°C. was nearly linear, whereas at 300 and 200°C. and at room temperature the linear relation was more evident with additions over two per cent. The rate of increase in strength was slightly higher with additions up to two per cent than with additions over 2 to 2-1/2 per cent. The outstanding superiority of magnesium over copper as well as the other elements might be attributed mainly to its higher solid solubility at 300 and 400°C. and its greater atomic diameter, which, judging from the favorable mechanical properties, seem to offset the effects of its higher diffusivity in aluminum. The elongations obtained with alloy No. 17 at 400°C. were erratic with respect to

both magnitude and reproducibility. The elongation dropped from 95 per cent at 300°C. to 46 per cent at 400°C. On examining the specimens after testing, however, there were definite indications that this behavior was caused by a weakening of the grain boundary phase. At 300°C., small cracks which followed the grain boundaries extended from the rupture back into the specimen, especially in the necked-down region. At 400°C. this effect was very pronounced. The ruptured section was very irregular and exhibited a definite grain boundary weakness. One specimen necked-down in four places within the initial 2-1/4 in. reduced section before rupturing in the weakest place. In a way this may be considered an indication of toughness for in spite of this unusual effect - the apparent weakening at the grain boundaries - alloy No. 17 at 400°C. was still far superior in tensile and yield strength to other alloys having a more favorable, finely crystalline fracture, transcrystalline or otherwise. As far as was determined in this investigation, additions of magnesium of the order of five per cent appeared to be most effective for improving elevated temperature properties for wrought aluminum alloys. The high yield strengths obtained at 200°C, as compared with those obtained at room temperature, for alloys 16 and 17 and also for alloys 1 and 2 of the copper group may be attributed to the fact that the alloys were in the overaged but not necessarily stress-free condition.

47. The results obtained with the Al-5Mg alloy in the early stages of the investigation gave a most promising outlook for magnesium as a basis for new aluminum alloys for elevated temperature applications. Although erratic elongations and highly irregular and undesirable grain boundary effects were obtained at 400°C. with single additions of magnesium, these effects were, to a great extent, checked by increasing the alloy content. In this respect alloy 42 improved the status of the 5 per cent magnesium addition. At this stage of the investigation, however, the work of Kempf, et al (1), of the Aluminum Company of America, featuring wrought polynary alloys containing 6% magnesium, was made available. This work did much to discourage the promising outlook for magnesium as far as the development of a general purpose alloy having better fabricating and service characteristics than "Y" alloy was concerned. It was pointed out, however, that the wrought 254 type alloy as well as the modified 177 casting alloy (Paragraph 13), possessed some qualities superior to those of "Y" alloy (at the expense of others) and as such were better suited as special purpose alloys. Kempf (1) stated that "it would appear that the state of the art of developing and utilizing light alloys at elevated temperature is at a stage where the metallurgists can be of maximum assistance only if the designer can define more accurately the specific combinations of properties required for an individual application". With this view in mind, it is not a remote possibility that a large part of the future market for alloys for elevated temperature applications may consist of special purpose alloys in preference to some of the standard alloys. For these reasons it is believed that further work with Al-Mg type alloys should be continued where Alcoa left off. It will be shown that the effects of elements such as tungsten, vanadium, molybdenum and

cerium should not be overlooked if such investigations were continued. These elements might prove instrumental, for example, in improving the resistance to creep of the 254 alloy without an otherwise disqualifying influence.

Additions of Cr

48. It is seen from the data in Table 3 that at room temperature chromium was as effective as manganese in increasing tensile strength in the range of 4-1/2 to 5-1/2 per cent. However, the strength-composition curves showed manganese to be more effective with additions less than 4 per cent. Chromium gave slightly better yield strengths than manganese for all compositions tested, although they were exceptionally low compared with the tensile strengths. The low crosshead speed probably accounted for this. At 300 and 400°C. it was evident from the general slope of the curves that little definite advantage would be gained by using over 3 per cent chromium. In fact, at 300 and 400°C. approximately 85 and 80 per cent, respectively, of the maximum effectiveness (over a range of 0.5 to 5 per cent Cr) was obtained with an addition of 2 per cent. Exceptionally low elongations at elevated temperatures were obtained with high percentages (2-5%) of chromium.

Additions of Co

49. The tensile data for alloys 22-24 clearly showed a considerable decrease in the rate of increase in tensile strength with additions of cobalt exceeding the eutectic (1 per cent) when tested at room temperature and 300°C. The increase in tensile strength obtained for the 1 and 4 per cent additions was 12 per cent at room temperature, 34 per cent at 300°C. and only 1.6 per cent at 400°C. The strength-composition curves indicated that at 300°C. additions of cobalt up to 4 per cent, perhaps more, would continue to be effective, whereas, at 400°C. little was to be gained, as far as improving tensile properties was concerned, with additions over 2 to 2-1/2 per cent. Undoubtedly when certain percentages of addition elements are exceeded, new phases lower the noted gain in mechanical properties. In comparing the curves plotted for the additions of cobalt with those of nickel it was found that at 400°C. the strengthening effect of each element was nearly the same with additions up to 3 per cent. Nickel was more effective in increasing the tensile strength in amounts over 3 per cent. At 300°C. however, the eutectic cobalt alloy gave a tensile strength equivalent to that of about 3-1/2 per cent nickel in aluminum. However, at both 300 and 400°C., the corresponding yield strengths of the nickel bearing alloys were somewhat higher. The data and curves suggested that nearly equal elongations would be obtained at 300 and 400°C. for binary alloys having equal additions of nickel and cobalt.

Additions of W and Mo

50. Although molybdenum proved considerably more effective than tungsten in increasing the tensile and yield strengths at room temperature,

it did not maintain its superiority at 300°C. At the latter temperature tungsten gave slightly better tensile and yield strengths with additions up to one per cent; over one per cent the effects of each element were about equal. At 400°C. one-half per cent additions of tungsten and molybdenum gave, as judged by the strength-composition curves, similar tensile and yield strengths, but a slightly lower elongation was obtained with the tungsten addition. In a range of 1 to 1-1/2 per cent, however, tungsten was considerably more effective in increasing the tensile strength but not the yield strength. At 400°C. the elongations of alloys 27 and 30, containing 4-1/2 per cent W and 2.8 per cent Mo, respectively, were nearly equal. In way of conclusion, at 300°C., increasing the amount of tungsten over a range from 1/2 to 5 per cent would increase the tensile strength appreciably, but no improvement in yield strength would be forthcoming with additions exceeding two per cent. At 400°C., the maximum effectiveness, as far as improving tensile strength was concerned, was obtained with about 1-1/2 per cent tungsten; a decrease in elongation and tensile-yield ratio was obtained with higher additions. With additions of molybdenum approximately 75 to 80 per cent of the highest strength reported (alloy 30, containing 2.8 per cent Mo) at 300 and 400°C. was obtained with additions of 1 to 1-1/2 per cent.

51. Mondolfo (18), in reviewing the work of Clark (19), reported the questionable solid solubility of 1.7 per cent W at the eutectic temperature which did not decrease with decreasing temperature. This may account for the increased effectiveness, over that of molybdenum, at both 300 and 400°C. of additions of tungsten in the order of 1 to 1-1/2 per cent. In fact, of the alloys containing various additions of Co, W, Mo, V, Ti and B, the 1/3 per cent W alloy (No. 26), gave the highest tensile strength at 400°C.; the yield strength obtained, however, was questionably low. Tungsten, in the above amount, also proved nearly as effective a strengthener as 2 per cent Cr and appreciably more effective than additions of the order of 5% Fe, 6% Ni, 11% Si, and 5% Cu. This seemed to support Mondolfo's belief that "if the solid solubility in aluminum is exact, tungsten as an alloying element for aluminum may be worth investigating".

Additions of V and Ti

52. At room temperature vanadium was about half way between molybdenum and tungsten with respect to its effectiveness in increasing tensile properties, while titanium was most effective of all, especially when present in relatively small amounts. The fine needle-like form of the compound $TiAl_3$ undoubtedly contributes to the strengthening effect of low percentages of titanium at room temperature as well as to grain refinement. On comparing alloys 32 and M-8, containing 0.9% V and 3.8% Ti, respectively, it is seen, at both 300 and 400°C., that approximately one-fourth as much vanadium as titanium gave equivalent tensile and yield strengths, although lower elongations were obtained with the titanium additions. Although alloy 33, with 2.7% V, was not tested at 300°C., the strength-composition curve extrapolated through that point seemed to indicate that additions of 1 to 1-1/2 per cent vanadium would give 85 to 90 per cent of the tensile and yield strength that would be obtained with

higher percentages. At 400°C., however, it was evident that additions of 0.15 to 0.30 per cent vanadium would give nearly 85 per cent of the over-all effectiveness obtained with additions exceeding 1 per cent. At 300°C. titanium might prove worthwhile with additions up to 2 per cent, whereas, at 400°C, it was seen that little improvement in tensile properties would be obtained with additions over 0.5 to 0.7 per cent, probably less. Vanadium has a solid solubility of 0.37 per cent at just below the eutectic temperature, which probably accounts partly for the high strengthening effect of small amounts, especially at 400°C.

Addition of B

53. Only one alloy (M-10) containing 0.44 per cent boron was tested. With additions of the same order of magnitude, titanium, vanadium, chromium and molybdenum proved more effective at 400°C. for increasing the tensile and yield strengths, while tungsten was about equally effective and cobalt less effective. The order of effectiveness of these elements was about the same at 300°C, except for molybdenum which gave somewhat lower tensile properties. Since boron had no pronounced favorable effects on the tensile properties, and considering other factors involved in the use of boron such as foundry characteristics and cost, the element could hardly be considered a feasible addition element to aluminum.

Additions of Ce

54. Next to magnesium, additions of cerium (alloys 35 and 36) gave the most unexpected and promising results of the investigation. The average tensile strength of these two alloys at 400°C. (2700 psi) compared favorably with those of alloys 19 and 21, containing 5.3% Mn and 4.9% Cr, respectively, and, as determined from the strength-composition curve for magnesium, the cerium addition would prove as effective as a 3.5 to 3.75 per cent addition of magnesium as far as improving tensile strength was concerned. The average elongation at the same temperature of alloys 35 and 36 (54%) was higher than that of either alloy 19 or 21; their average yield strength (1820 psi) however, was lower than that of any of the three alloys compared. Although both alloys 35 and 36 contained a total of other rare earths, mainly lanthanum, equal to the cerium content, it is highly probable that for all practical purposes the total rare earth content may be considered as cerium. Smirnov (22) states that "diagrams showing properties of aluminum alloys of rare earths group are analogous to diagrams showing properties of the Al-Ce system. Likewise, the behavior of these metals (La, Pr, Nd, Tb, Sm, Eu) in aluminum is analogous to the behavior of cerium in them."

55. One of the outstanding characteristics of alloys 35 and 36 was their good formability, being considerably better than for the other alloys having high alloy contents, with the exception of the eutectic silicon type. This implies that, from the standpoint of both formability and strength, appreciable amounts of other addition elements could,

as with the eutectic Al-Si alloy, probably be used along with a comparatively high total rare earth content, say from 8 to 13 per cent. The eutectic of the Al-Ce system is at 13 per cent cerium, and, for reasons previously mentioned, a total of 13 per cent rare earths, at least half of which is cerium, might give an eutectic-type structure. "Misch-metal" containing at least 50 per cent cerium, with low iron content, is available and, therefore, it is believed that the effects of various addition elements on the eutectic or nearly eutectic Al-Ce alloy* should be thoroughly investigated for elevated temperature applications of both a general and specific nature. No such work has been reported in the literature to the author's knowledge. The additional expense involved in the use of mischmetal could be tolerated at least for special applications if the alloy or alloys met the prescribed qualifications for a given job. Although the ingot cost might be doubled, the percentage increase in cost in the fabricated condition would, on the basis of present marketing practice, be considerably less. Moreover, a decrease in cost could be expected with an increase in demand for this element.

56. Favorable properties have been reported within the past few years for the Mg-Ce type alloys; it is believed that further investigations should be continued along these lines, with special emphasis placed on the effects of the more unusual addition elements discussed herein (V, W, Co) as well as of Cu, Al, Mn, Cr and Zn.

57. A summary of the amounts of single addition of each element required to give optimum tensile properties at 300 and 400°C., in order of decreasing effectiveness, is given in Table 6. The values given in each case for the upper limit would give the optimum strengthening effect for wrought binary alloys, whereas the lower limits would be more feasible for polynary alloys, except in the case of magnesium and cerium. As far as increasing strength at elevated temperatures was concerned, these data showed the feasibility of using tungsten, vanadium and molybdenum in more than just grain refining amounts. That the order of effectiveness, as well as the amount required for optimum effectiveness, for some elements changed in increasing the testing temperature from 300 to 400°C., did much to clarify the fundamental behavior of these various elements in aluminum at elevated temperatures. At some critical temperature between 300 and 400°C., the real effect of the respective addition elements in strengthening the matrix came into play. The critical temperature would undoubtedly vary with different alloys; it would be reasonably valid to assume that for the composition ranges prescribed in Table 6, the higher up in the 400°C. column the element was, the higher would be the critical temperature. The most marked change of positions obtained in increasing the testing temperature from 300 to 400°C. was found with cobalt and manganese - the former

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*Note: Hereafter, cerium content refers to high purity mischmetal containing 99% rare earths of which 50.6% is cerium.

dropping from the sixth position at 300°C. to the thirteenth at 400°C., and the latter rising from the eleventh to the fifth position, respectively.

Eutectic Alloys

58. Four eutectic or nearly eutectic alloys were included in this investigation, and by extrapolation from the strength-composition curve a reasonably indicative picture of the tensile properties of a fifth, containing cerium plus other rare earths, was obtained. The tensile properties of these five alloys, and for comparison that of alloy M, containing 5% Mg, have been repeated in the following table:

BINARY EUTECTIC ALLOYS

Alloy No.	Percent Addition	Tensile Strength p.s.i.		Yield Strength psi, 0.2% offset		Elong. in 2 in, Percent	
		300°C.	400°C.	300°C.	400°C.	300°C.	400°C.
8	Si 10.78	4180	1760	3140	1190	60.5	69
11	Ni 5.88	5400	2270	3600	1370	43	48
14	Fe 1.58	3400	1930	2430	1330	43	45
23	Co 1.05	3650	1260	1960	930	60	50
X	Ce 6.5 R.E. 6.5	-	3200*	-	2300*	-	45*

*Extrapolated values.
Crosshead speeds and heat treatments given in Table 3.

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The high decrease in tensile and yield strengths obtained even for the eutectic alloys in testing at 400°C. is clearly evident. The critical temperature accounting for most of the decrease of these as well as most of the other alloys investigated seemed to be in the range of 330 to 360°C. The feasibility of using appreciable amounts of the rare earths, as compared with magnesium, is indicated.

Polynary Alloys

59. Only six polynary experimental alloys were included in this investigation (alloys 37-42). In comparing the tensile properties of alloys 37 and 38 with those of alloy 17, it is evident that additions of silicon will definitely decrease the tensile and yield strengths, which in turn means that the Mg_2Si compound formed was less effective as a strengthener at elevated temperatures than the magnesium it displaced from solid solution. The properties of alloy 39 showed that the tensile strength, but not the yield strength of an alloy of the Al-Mg-Si type, would be decreased further by small additions of grain refiners such as

titanium and/or cerium. Although alloys 40-42 also contained silicon, the other addition elements improved the tensile properties over those of alloy 37; in fact, alloy 42, containing Mg, Mn, Co, Fe and Si additions, had slightly higher strength than alloy 17 at 400°C., and thus had the best tensile properties found in this investigation. The magnesium recovery for this alloy was 1-1/2 per cent low - had this not been the case the tensile and yield strengths at 400°C. undoubtedly would have been appreciably higher.

60. Furthermore, the ruptured surfaces of sheet specimens of alloy 42, and also of alloys 38 and 39, after stabilizing and testing at 400°C., and although seemingly of an intercrystalline type, showed a finer crystalline structure, and a somewhat more brittle fracture, in contrast to the very irregular, intergranular type of fracture obtained with alloy 17. Alloys 37, 40 and 41, however, exhibited a fracture more irregular than those mentioned above but still were not as bad as that obtained for the straight Al-5Mg alloy. Also the erratic effects on elongation were not found in testing alloys 37-42. From these observations it appeared that in testing at 400°C. only minor alloy additions would be required to offset the grain boundary effect, which seemed to be inherent with the Al-5Mg alloy, as far as eliminating the erratic elongations was concerned. It was also evident that the more uniform and fine grained fracture of polynary alloys containing in the order of 5% Mg was indicative of greater strength only when such a structure was obtained with a fairly high alloy content, as in alloy 42, and not with alloys yielding equally attractive hot fractures but having comparatively low alloy content, e.g., alloys 38 and 39.

<u>Plates 3, 4 and 5.</u>

61. For the most part the effects of each element on the tensile properties at elevated temperatures have been discussed individually; comparisons were made in some cases with those elements having similar metallurgical characteristics. No attempt has been made, however, to make an overall evaluation of all the elements with respect to their relative effects on the tensile properties of aluminum at elevated temperatures.

62. A rapid survey of tensile properties at 400°C. of the outstanding binary alloys investigated can be found in <u>Plate 3</u>. The superiority of alloy 17 is easily seen. It was interesting to note that cerium occupied two of the first five positions. The relative positions of alloys 26, 33, 20, 11 and 30 suggested that nickel might be partially or totally replaced to advantage in some alloys by tungsten, vanadium, chromium and molybdenum, as far as strength was concerned. Titanium (M-8) proved to be less of a strengthener than the other grain refining elements even when present in a comparatively high amount. Copper, by itself, did not come up to expectations. It is obvious that in commercial alloys containing 4 to 8 per cent copper at least 50 per cent of the strengthening must come from other addition elements if the alloys are to be used in a temperature range of 350 - 400°C. The best tensile-yield ratios were obtained with alloys 21, 20, 32 and 34.

63. Since the percentages of addition elements for a number of alloys on Plate 3 were somewhat high compared to the amounts in which they are more apt to be used, two more groups of curves were plotted on Plates 4 and 5, showing the effects of the addition elements, sectioned at 1-1/2 and 4 per cent, on tensile properties at both 300 and 400°C. The tensile and yield strengths were taken from the strength-composition curves plotted from the data in Table 3. In each case the elements were arranged in order of their decreasing effectiveness on tensile strength. It was obvious that the order of effectiveness changed with both composition and temperature of testing. With 4 per cent additions magnesium maintained its superiority at both 300 and 400°C, whereas, with 1-1/2 per cent additions tungsten and chromium proved slightly more effective than and cerium nearly as effective as magnesium at 400°C.; at 300°C, vanadium and titanium were slightly more effective than magnesium in improving tensile strength. However, the 1-1/2 per cent additions of the above elements at either temperature did little to improve the yield strength over that of the magnesium addition; in fact, at 400°C, tungsten gave an eratically low yield strength. At both 300 and 400°C., with the 4 per cent additions, it was interesting to note the relative positions of Mo, Mn, Ni, Co and Fe, both with respect to each other and to the other elements. The good strengthening effect shown by cobalt at 300°C. was greatly decreased at 400°C. Nickel was lower in effectiveness at each temperature than would be expected, considering its seemingly important present position in aluminum alloys for elevated temperature use. At 400°C., 4 per cent molybdenum proved as effective a strengthener as an equal amount of manganese.

64. Plates 3, 4, and 5 contain data requiring careful interpretation. In considering the relative effectiveness on tensile properties at 300 and 400°C. of single additions of Ce, Cr, V, Mo, W and Ti as compared with the more common additions of Cu, Mn, Mg, Ni, Fe, Co and Si, an important question may be raised. Do the favorable results obtained with the former group serve only to minimize the significance of the short-time tensile test as a criterion for predicting what can be expected of a given element, as well as to re-emphasize the importance of other service characteristics such as thermal conductivity, resistance to creep and fatigue, or, conversely, do they point out the possibilities, which have not been investigated (at least not reported) of these elements as being valuable additions to aluminum alloys for elevated temperature applications, especially when present in amounts greater than normally would be used, i.e., Ti, Ce or V as grain refiners only? From the experimental results obtained it is believed that, in spite of the inherent foundry difficulties, sizable additions of elements such as V, Mo, W and Ti should not be overlooked in the development of general or special purpose aluminum alloys for elevated temperature applications. Future research and development on foundry techniques, especially with chill and permanent molds, may provide new and better methods to cope with thess high melting point addition elements.

65. An interesting ramification of the tensile data obtained at 300 and 400°C. of 25 selected alloys, including all the alloys in Plate 3,

can be found in Table 4. The percentage decreases in tensile and yield strengths obtained for 300 and 400°C. were determined and the alloys, in turn, arranged in order of increasing percentage difference in tensile strength. The corresponding order of their tensile and yield strengths at 400°C. are given in the right hand columns. One of the most obvious things to be noted is the comparatively high decrease in strength (approximately 50%) obtained for alloys having good tensile properties at 400°C, namely, alloys 17, 35, 36 and 33. But more interesting factors or information are suggested in Table 4. In considering the low crosshead speed used up to the yield loads (0.05 in. per min.), the rate of decrease in yield strength in going from 300 to 400°C., as well as the actual yield strength obtained at the higher temperature, could be indicative of the relative resistance to creep for additions of certain elements, provided the tensile strength was of a reasonably high order to begin with. Thus, of the first 12 alloys in the Table (those above the dotted line) having the lower rates of decrease of tensile strength, nine of the alloys, namely, 59, 19, 31, 56, 14, 30, 21, 20 and 27, also had low rates of decrease in yield strength. Of these nine alloys, seven - Nos. 59, 19, 31, 30, 21, 20 and 27, had relatively high tensile and yield strengths at 400°C. Hence, it is highly probable that additions, above certain critical amounts in each case, of Cr, Mn, V, Mo and W would be instrumental in improving the resistance to creep of aluminum alloys at elevated temperatures, even though some are essentially grain refiners as minor additions. With a similar line of reasoning, the effects, as far as improving resistance to creep is concerned, of single additions of Mg, Ni, Ce and Co would not be marked; Fe and Cu, on the other hand, should prove beneficial with relatively high amounts. The foregoing observations have been made fully realizing the limitations of predicting the behavior of elements in polynary alloys from that found in binary alloys.

Macrostructures - As Cast

66. Specimens of the as-cast alloys were macro-etched for grain size. These have been reproduced in Plates 6, 7, 8 and 9. At nearly constant cooling rates, it was found that appreciable quantities (approximately 3% or more) of iron, manganese, chromium, cobalt, titanium and, to a lesser extent, nickel, produced a columnar, dentritic type of structure, progressing inwardly from the mold wall to the center of the cast section; nickel produced a structure more fibrous than columnar. The above type of structure tended to make hot working difficult, especially with high percentages (over 4 to 5%) of manganese and chromium; however, these alloys could be rolled provided that small reductions of area were made between intermediate anneals. In general, in rolling to a reduction of area over 85 to 90 per cent, some metal loss due to edge cracks is to be expected. Addition up to 5 per cent of magnesium and copper, to 1 per cent of cobalt and iron, from 2 to 3 per cent of molybdenum and less than 2-1/2 per cent nickel, produced a normal type solidification, characterised by a columnar or fibrous at the edges followed by coarse to fine equiaxed crystals as the center of the cast section was approached.

Ordinarily such structures can be hot worked without difficulty, but special care would be required for alloys containing 5 to 6 per cent magnesium or over 7 to 8 per cent copper. These percentages are about the optimum values useable for wrought aluminum alloys.

67. Tungsten showed a marked grain refining ability with additions up to nearly 3 per cent, whereas, vanadium and titanium were most effective up to 1 per cent. Alloy 33, containing 2.7% V, had a somewhat coarser grain size than alloy 32, containing 1% V. Alloy M-8, containing 3.8% Ti, had an interesting, eutectic silicon-type structure, characterized by the high melting point, needle-like constituent evenly dispersed throughout the matrix. This structure, however, did not come up to expectations as far as its strength at elevated temperatures was concerned. Alloy M-10, containing 0.44% B, was also very fine grained. Molybdenum, in the three compositions noted, proved to be less of a grain refiner than the most effective amounts of the elements discussed above. Alloys containing these elements could be hot worked satisfactorily.

68. Both macro examinations and chemical analysis of various parts of the castings indicated that special precautions would be necessary to prevent excessive segregation of constituents in alloys containing appreciable amounts of titanium, molybdenum, tungsten, boron, vanadium, and, to a lesser extent, cobalt. This situation stemmed from the fact that in working with alloys containing relatively high percentages, i.e., higher than ordinarily would be used for elements of this type, it was of course impossible to pour at temperatures at which any of the compounds would be completely in solution and at the same time obtain a sound, gas-free casting. Satisfactory castings could be made in most cases, however, if adequate stirring of heats before pouring and sufficiently rapid cooling rates were used. In this investigation most trouble was encountered in making alloys 27, 30, M-8 and 24.

Fracture Characteristics - As Cast

69. Fractures were made of 1-3/4 in. x 1-3/4 in. x 1/2 in. thick specimens cut from the permanent mold castings, each adjacent to the part from which the billet was cut to prepare tensile specimens. Each specimen was saw notched approximately 3/32 in. deep across the 1-3/4 in. x 1-3/4 in. face to localize fracture stresses. The fracture characteristics were evaluated mainly with respect to their relative brittleness or ductility, as determined by the angle of bend before fracture, toughness and impact strength, as indicated by the number of equally hard blows necessary to cause fracture, and type of cast structure. Toughness was considered a factor only with those alloys which gave brittle type fractures for it would naturally take more blows to fracture a ductile alloy which in this case could hardly be considered a criterion for toughness.

70. The fractures on Plate 10 are typical for the most part of the various types obtained in this investigation, embracing the very ductile type of the eutectic Al-Fe alloy (No. 13) to the very brittle

Al-5Fe type (No. 15). The granular type structure, in which the grains themselves seemed to stand out, with no apparent matrix, was not shown in Plate 10. Alloys 2 and 17, containing 3% Cu and 5% Mg, respectively, gave typical fractures of solid solution alloys having an irregular, non-granular appearance. Both were moderately brittle, but alloy 17 was much tougher and broke as though the grain boundary phase were exceptionally strong. Alloy 8, containing 11% Si, had a brittle, moderately tough fracture; its structure was dense and non-granular, having a grayish color resembling ordinary cement. The 0.6% Si alloy (No. 6) gave a very irregular, loosely fibrous structure which seemingly indicated a weakness or upsetting at the grain boundries. The 5% Fe alloy was very brittle and obviously had low impact strength. Its structure consisted of particles of Fe_2Al_7 compound evenly dispersed in a dark gray matrix. An examination of the fracture of alloy 11, containing 6% Ni, revealed much as to why this element has been an effective strengthener at elevated temperatures. This alloy gave a tough, moderately brittle break; its structure - being dense and fibrous near the edges, though "woody" near the center - consisted of long, interlocking fibers or needles of the $NiAl_3$ compound. Little matrix was apparent. Such a structure would undoubtedly be condusive to high hot strength.

71. Alloys 22, 25, 31, and, to a lesser extent, 26 and 28 gave fractures similar to that of alloy 13, shown on Plate 10. The fracture of alloy 19, containing 5% Mn, also consisted of a mass of interlocked needles ($MnAl_6$), but this effect was not quite as pronounced as with alloy 11, mainly because more matrix (eutectic) was visible. Alloys 20 and 21, containing 2 and 5% Cr, respectively, gave a brittle type fracture with the former being slightly tougher. Each had a dense, amorphus appearance - neither granular nor fibrous, although in the Al-5Cr alloy some $CrAl_7$ compound was evident throughout the matrix. Alloys 27, 29, 30, 32, M-8 and M-10 gave brittle type fractures, each characterized, with the exception of alloy 29, by a dense, fine grained matrix with compounds of their respective addition elements evenly dispersed throughout. Alloy 29 had a brittle but moderately tough fracture with a somewhat fibrous structure. Both alloys 35 and 36, containing cerium, gave moderately brittle and tough fractures, having a dense, fine grained structure. No definite compounds were visible. All of the polynary alloys (37-42), featuring 4 to 5% Mg, had in common the brittle but moderately tough type of fracture with a dense, fine grained structure. Each fracture had a bluish coloration which could not be accounted for. Alloy 42, the strongest made in this investigation, had, in addition to the foregoing qualities, a fine, needle-like compound, probably Co_2Al_9, evenly dispersed throughout its matrix. This structure looked good for elevated temperature applications. In general, some relationship seemed to exist between cast structure (fracture characteristics) and hot strength in the wrought condition, for those alloys having a brittle type fracture, with varying degrees of toughness, exhibited, in most cases, superior hot strength in the

wrought condition; the relative hot strengths seemed related to the corresponding degree of toughness in the cast state. More experimental data would be necessary, however, to substantiate this.

Hot Rupture Characteristics

72. In recent years considerable importance has been placed on the type of fractures or ruptures obtained when testing various metals and alloys in tension at elevated temperatures. Such observations might contribute materially to a better understanding of the behavior of some materials at high temperatures, especially with respect to their indicating the weakest phase of a given alloy system. The latter could be done quantitatively by determining, by means of a microscope if necessary, whether the alloy broke in a transcrystalline or intercrystalline manner or perhaps both, depending on the orientation of some of the grains with respect to the direction of the applied stress at the ultimate load. Only visual and thus, for the most part, qualitative examinations were made in this investigation. The reductions in area of the strip specimens tested were used as a criteria of the relative brittleness of the alloys at elevated temperatures. The alloys were evaluated in this respect as follows:

<u>Brittle</u> - 0 to 10% reduction in area (sight estimate)
<u>Moderately brittle</u> - 10 - 40% R. A.
<u>Ductile</u> - 40 - 75% R. A.
<u>Very Ductile</u> - 75 - 99% R. A.

Other characteristics of the ruptured cross-sections fell in one of the following categories:-

(1) Fractured surface appeared dense, finely crystalline, smooth-fracture probably transcrystalline.
(2) Fractured surface appeared coarser grained than above and not as smooth; was difficult to ascertain by sight whether fracture was more apt to be intercrystalline or transcrystalline - probably both.
(3) Fractured surface irregular and grainy with definite indications of grain boundary weakness in some cases - fracture probably intercrystalline.

Of course in order to make an evaluation of one of the above, it was necessary that the alloy be brittle enough (small enough reduction in area) to reveal a representative section of metal. A good many specimens necked-down to the extent that it was impossible to make any evaluation outside of their being highly ductile.

73. With the exception of alloys 2, 6, 25, 28 and 31, which were ductile, all of the alloys exhibited a brittle or moderately brittle rupture at room temperature; alloys 8, 10, 11, 23, 30 and 35-42 had a transcrystalline type fractured surface, as described under (1) in the preceding paragraph, whereas the remaining alloys followed the description

given in (2). Of the alloys tested at 200°C, Nos. 7, 9, 103 and 16 became ductile at that temperature and of course very ductile at 300 and 400°C. Of most interest were the fractures obtained at 300 and 400°C. At 300°C, alloys 4, 8, 10 and 11 became ductile, having lost the brittleness shown at 200°C.; alloys 22, 23 and 56 were also ductile at 300°C., and probably were ductile at 200°C., although they were not tested at the latter temperature. At 300°C., alloys 4, 8, and 11, though somewhat ductile, exhibited a type of fracture described under (2) above. Alloys 19, 21, 24, 27, 30 and M-8 were still brittle at 300°C., whereas alloys 15, 17, 60, 20, 26, 29, 32 and M-10 were moderately brittle. Each of the alloys of the latter two groups had the appearance given in (2), except alloys 15 and 17 which had the intercrystalline type fracture described under (3) in the preceding paragraph. Alloys 33-42 were not tested at 300°C., but, judging from the results obtained at 400°C., alloys 34 and 35 were probably ductile and the others moderately brittle.

74. At 400°C., alloys 19 and 21, containing 5% Mn and Cr, respectively, remained in a brittle category, while alloys 15, 17, 20, 24, 27, 29, 30, 32, M-8 and 34-42 were moderately brittle. Of these, alloys 15, 17, 37, 40 and 41 gave fractures seemingly of the intercrystalline type, whereas the others were as described in (2), Paragraph 72. One of the interesting things noted in this study was that, of the 19 alloys having comparatively high strength at 400°C.- namely, alloys 11, 15, 17, 19, 20 21, 26, 27, 30, 33, M-8, 35, 36, and 37-42, alloys 11, 26, 33, 35 and 36, containing 6% Ni, 1.3% W, 2.7% V, 6% Ce and 8.6% Ce, respectively, gave either a ductile or very ductile type of rupture; the others were brittle or moderately brittle. Perhaps the dense, fibrous cast structure described in Paragraph 69, would account for this behavior in alloy 11 - having the strength to give a high ultimate load and the tenacity to withstand a high reduction in area. As was intimated in Paragraph 55, the Al-Ce system was somewhat analogous to the Al-Si system, especially with respect to hot formability and cast structure, and, therefore, the high ductility at elevated temperature of the Al-Ce system might justifiably be compared with that of the eutectic Al-Si alloy which also gave a high reduction in area at 400°C. but, surprisingly, a lower strength. The behavior of alloys 26 and 33 could not be accounted for. The net results of these observations on hot rupture characteristics of binary aluminum alloys indicated that:

(1) The intergranular type of hot fracture should not necessarily be considered as indicative of an undesirable type of structure for elevated temperature applications, especially where additions of at least 4-5% magnesium are concerned.

(2) Generally, a brittle or moderately brittle type rupture, as evaluated in Paragraph 72, was indicative of comparatively high hot strength but, conversely, a ductile type rupture could not always be associated with low hot strength, especially with substantial additions of nickel or the rare earths.

Microstructure

75. Specimens in the as-cast condition, from at least one alloy of each of the series tested, were set aside for metallographic examination. In addition, in order to determine the effects of hot forming and stabilization on the structures - especially with respect to the effects on distribution, size and form of the constituents, specimens were taken from the strip tensile specimen tested at room temperature and from specimens of the same alloys tested at elevated temperatures after a prolonged stabilization at the test temperature. Unfortunately, the examination was not completed in time to be discussed in this report. All specimens in the as-cast condition from the binary alloys were completed, except those containing cerium (alloys 35 and 36), whereas, the metallographic work for the specimens in the wrought (rolled) condition was completed only for alloys 1 - 17, inclusive.

76. Photomicrographs of as-cast structures of only six alloys containing the more unusual addition elements were included in the report (Plates 11 and 12). The other structures, being comparatively common, have been covered by Mondolpho (17). Of this group boron and tungsten formed needle-like compounds with aluminum (Figs. 1 and 2, Plate 11).. In considering the nature of its compound, boron probably would have had a considerably higher strengthening effect at elevated temperatures than reported, had it been present in a greater amount. The difficulty in handling this element, however, would hardly warrant its use. Likewise, the form, size and distribution of the compound formed in the Al-5W alloy seemed conducive to higher strengths at 300 and 400°C. than reported. The oxides present in this case were later removed, for the most part, by remelting and stirring. It was evident that considerable care must be taken to produce sound metal with high amounts of tungsten, but not to the disqualifying extent found in working with relatively small additions of boron. It is believed that the use of tungsten, even in the order of 2 to 4 per cent, in aluminum alloys for elevated temperature applications should be thoroughly investigated. The compounds formed with appreciable amounts of Mo (Fig. 4, Plate 11), Ti and Co (Figs. 1 and 2, Plate 13) were somewhat angular compared to that formed with V (Fig. 3, Plate 11). Vanadium however, has some solubility in aluminum which may account partly for the rounded form of the VAl_7 compound.

77. Figure 2, Plate 13 was interesting in that it showed the desirable effects of fabrication on the breaking-up and redistribution of the large primary crystals of Fe_2Al_7 compound shown in the as-cast condition in Fig. 1. It showed further that the stabilizing treatment at 410°C. did not tend to change the form, e.g., by rounding, of the compound. Although the metallography was not completed for the other wrought alloys having large primary crystals in the cast state, for example, those discussed in the preceding paragraph, there is no reason to believe that the effects of hot forming by forging and rolling should not have been equally desirable.

78. Macrostructures of the wrought Al-5Mg and Al-6Ni alloys on Plate 14 were of special interest in that they revealed the effects of the stabilizing treatment at 410°C. on the as-rolled structures. The Al-5Mg alloy had completely recrystallized which, in conformance with the discussion in Paragraph 17, indicated that this type of structure was not necessarily detrimental to hot strength and perhaps resistance to creep, especially for alloys containing 4 to 6 per cent magnesium. Stabilization of the same alloy for 11 days at 310°C. completely obliterated the as-rolled structure and started nucleation, but no appreciable grain growth took place. As was evident in Fig. 4, the dense, fibrous structure of the Al-6Ni alloys was not easily recrystallized. Some nucleation occurred, but complete recrystallization, at 400°C. of alloys containing 4 to 6 per cent nickel would, if possible, undoubtedly require a comparatively long time. Macro-examination also showed that the Al-5Fe alloy was not subject to recrystallization during stabilization at 410°C. for 16 days, whereas the same treatment had recrystallized, but apparently with a much lower rate of growth compared to that of the Al-5Mg alloy, the as-rolled structure of the Al-1.6Fe alloy. The effects on the recrystallization of aluminum of the more unusual elements covered in this report would be of interest, but this work has yet to be completed.

Hardness

79. A general discussion of the methods used for hardness determinations was given in Paragraph 36. The results are given in Table 5. No attempt was made in this investigation to make correlations of room temperature hardness with other properties. As would be expected, however, the room temperature hardness, obtained for each series of alloys, either in the cast or wrought condition, increased with increasing alloy content - the highest values being obtained with alloys containing approximately 9% Cu, 11% Si, 6% Ni, 5% Mg, 10% Mg, 5% Mn, 8% rare earths, and the 6 polynary alloys featuring 4 to 5% Mg, (alloy Nos. 4, 8, 11, 17, 18B, 19, 36 and 37-42, respectively). It was interesting to note that of the above group of alloys all but Nos. 4 and 8 (the Al-10Mg alloy was not tested for other than hardness) gave relatively high strengths at 400°C., as compared with the other alloys tested. The "recovery hardness" values for the wrought alloys served mainly to indicate the relative effects of the different stabilizing treatments on hardness, which in turn would give an indication of diffusion or structural changes that may have occurred during the heat treatment. The most significant and uniform results are obtained by reading the columns vertically, giving the change in room temperature hardness with change in composition after treatment at a given temperature. In reading the columns of the table from left to right to compare the hardness values obtained for a given alloy after stabilizing and testing at different temperatures, some of the results were too erratic to be truly significant and no more than a trend may be indicated. The increase or decrease in RH hardness obtained in some cases were not of the same order of magnitude as those obtained with the RE scale. One factor that might have accounted for this behavior was the relative thickness of the oxide film on the specimens which was found to have some effect on the hardness values, especially with the RH scale.

SUMMARY AND CONCLUSIONS

80. As a preliminary part of an overall investigation to develop new aluminum alloys for elevated temperature applications, especially within the range of 300 to 400°C., the effects on short-time tensile properties at elevated temperatures, after sufficiently long stabilization treatments, of additions to aluminum of various percentages of 14 elements were determined. Forty-three binary and six polynary wrought alloys were investigated.

81. The relative strengthening effect of each element at 300 and 400°C., in amounts which gave optimum tensile properties under the prescribed test conditions, can best be found in Table 6. The relative effectiveness of amounts other than those given in the latter table can be found in Plates 3 - 5. One must use discretion, however, in evaluating experimental data of binary alloys with the idea in mind of determining the overall suitability of certain elements for use in polynary alloys, especially for elevated temperature applications. The amounts which may give the highest strenthening as single additions may or may not be feasible, depending on the particular application, from the standpoint of effect on other properties such as formability, thermal conductivity and density. Hence, no attempt will be made to recommend compositions, on the basis of the data obtained in this investigation, that might prove better than standard alloys now in use. To do so, without at least a systematic investigation of ternary and quaternary alloys, would be meaningless. The compositions on Table 6 are given fully realizing that hot strength is only one of several characteristics required of an alloy for use at elevated temperatures. But hot strength is one of the most important and more readily determined of the significant properties and, thus, offers a most logical stepping stone to an overall evaluation of the addition elements.

82. Of the fourteen elements investigated, single additions of magnesium and cerium in the order of 5 and 8 per cent, respectively, proved to be the most effective strengtheners at 400°C. At the same temperature, however, additions of 5 to 8 per cent copper and 11 per cent silicon were surprisingly ineffective.

83. In general the order of optimum effectiveness of the element in improving tensile properties was, as is shown in Table 6, a function of both composition and temperature. On comparing nickel and manganese it was indicated that additions of nickel over 3.5 to 4 per cent were more effective than equal amounts of manganese, whereas, with additions less than approximately 3 per cent manganese was more effective. At the same temperature additions of cobalt up to 2 per cent gave greater strength than additions of nickel up to about 4 per cent, and proved superior to manganese at all compositions tested. At 400°C, however, cobalt did not show this superiority, and manganese proved most effective of these three elements for all compositions tested. At 400°C. single additions of Mg, Cr and Ce only proved more effective than Mn.

84. Of the elements V, Ti, W, Mo, Cr and B, vanadium was most effective as a strengthener, in prescribed amounts, at 300°C, but in turn lost this superiority to Cr and W at 400°C. Further, at the latter temperature additions of W of about 1.5 per cent not only proved as effective a strengthener as 2% Cr, but appreciably more so than additions in the order of 5% Fe, 6% Ni, 11% Si and 5% Cu. Vanadium was surprisingly effective at 400°C. in amounts less than 0.6 per cent. It was shown that the use of W, V and Mo, in more than just grain refining amounts, should not be overlooked in the development of new aluminum alloys for use at elevated temperatures, in spite of the foundry difficulties inherent in the use of these elements.

85. Other conclusions of secondary importance were arrived at through the study of other than tensile properties at elevated temperatures. These have been summarized below:

(1) At nearly constant cooling rates, it was found that appreciable quantities (approximately 3% or more) of Fe, Mn, Cr, Co or Ti and, to a lesser extent, Ni, produced a columnar, dentritic type of permanent mold cast structure; nickel produced a structure appearing more fibrous than columnar. This type of structure tended to make hot working difficult, especially with high percentages (over 4-5% of Mn and Cr.

(2) Additions up to 5% of Mg and Cu, to 1% of Co and Fe, from 2 to 3# of Mo and Less than 2-1/2% Ni, produced a normal type solidification. Ordinarily such structures could be hot worked without difficulty, but special care would be required for alloys containing 5 to 6% Mg or over 7 to 8% Cu. Binary alloys containing W, V, Ti, Mo and B could be readily hot worked in the amounts tested.

(3) An examination of the Al-6Ni alloy revealed much as to why nickel was conducive to high hot strength. This alloy gave a tough, moderately brittle break; its structure (being dense and fibrous near the edges, though "woody" near the center) consisted of long, interlocking fibers of the $NiAl_3$ compound. Little matrix was apparent. The same effect was found in the Al-5 Mn alloy but was not as pronounced. It was evident that the mechanism which accounted for the superiority of mangenese over nickel, as a strengthener at 400°C., was more fundamental in nature and may probably be attributed to the fact that manganese has some solubility in aluminum whereas nickel does not.

(4) A general relationship seemed to exist between fracture characteristics in the cast state and hot strength in

the wrought condition, for those alloys having a brittle type fracture, with varying degrees of toughness, exhibited, in most cases, superior hot strength; the relative hot strengths seemed related to the corresponding degree of toughness in the cast state. More experimental work would be required to substantiate this.

(5) Observations of the hot rupture characteristics of wrought binary aluminum alloys indicated that (1) the intergranular type of hot rupture should not necessarily be considered as indicative of an undesirable type of structure for elevated temperature applications, especially where additions of 4 to 5% Mg were concerned and (2) generally a brittle or moderately brittle rupture was indicative of comparatively high hot strength but, conversely, a ductile type rupture could not always be associated with low hot strength, especially with substantial additions of nickel or the rare earths.

(6) The relatively high strength obtained for the Al-5 Mg alloy at 400°C. indicated that a completely recrystallized structure was not necessarily detrimental to hot strength; that is, such a structure would, at least in some cases be as strong as one inherently large grained as a result of its metallurgical history other than best treatment.

RECOMMENDATIONS

86. From the results obtained and observations made in this investigation on the development of new aluminum alloys for elevated temperature applications, it is recommended that:

(1) A thorough, systematic investigation be undertaken on not only binary alloys of aluminum and cerium, but also on ternary and quaternary alloys of this system. The effects of additions of Mg or Cu, Mn, Ni, W, V, Cr and Co, singly and collectively, should be emphasized. These elements should be used, insofar as possible, in the amounts given in Table 6. This investigation should be undertaken with the idea in mind of developing casting and wrought alloys for both general purpose (expense permitting) and special purpose applications.

(2) In line with the above investigation, the effects of replacing part of the silicon content of the 32S and A132 standard alloys with cerium, as well as of the eutectiferous Al-Si experimental alloy containing 4 to 6 percent Cu (Paragraph 14), be investigated.

(3) Further investigations be continued on wrought alloys containing 4-6 percent Mg and casting alloys containing approximately 10 percent Mg. This investigation should embrace ternary and quaternary alloys featuring appreciable amounts of Ce, W, Mo and V, and also similar additions to the 254 and modified 177 type alloys developed at the Aluminum Company of America (Paragraph 13). The latter investigation should be concerned with alloys for specific applications at elevated temperatures.

(4) The Mg-Ce and effects of Al, Cu, Mn, Ni, W, V, Mo and Zn on this system be investigated further.

(5) In future investigations, the stress-rupture test is preferred to the short-time tensile test in evaluating certain polynary alloys. Experimental alloys having outstanding stress-rupture properties should then be subjected to creep tests.

BIBLIOGRAPHY

1. L. W. Kempf, W. A. Dean, R. B. Wrege: "Investigation of Aluminum Alloys for Application at Elevated Temperatures", Cleveland Research Division, Aluminum Research Laboratories, Aluminum Company of America, Report No. 4-45-6, 01-257 and 01-256, September 12, 1945.

2. Zay Jeffries, Chairman, Metals Conservation and Substitution Group - The Advisory Committee on Metals and Minerals: "Report on Beryllium", Metals and Substitution Group Report No. 63 to the War Production Board, September 1942.

3. W. G. Thompson, et al, "Final Report on Beryllium-aluminum Alloys for Engine Parts". OSRD No. 1656, Serial M-107, July 29, 1943. Restricted.

4. E. I. Brimelow: "New Aluminum Alloys for Use at Elevated Temperatures", Royal Aircraft Establishment, Farnborough, England; Report M.4169B, September, 1942. Controlled distribution.

5. E. I. Brimelow: "Note on Chill-Cast Components in Aluminum Alloy R. A. E. 55, Made for "Ricardo" Experimental Engine Units", Royal Aircraft Establishment, Farnborough, England; Note M.7155, August, 1943. Controlled distribution.

6. E. I. Brimelow: "Hercules pistons in R.A.E. 40C and R.A.E. 47D Alloys Examined After Single Engine Tests", Royal Aircraft Establishment; Report No. M.4169T, April, 1944. Controlled distribution.

7. "Investigation of the Mechanical and Physical Tests of Some Wrought and Cast Aluminum Alloys Carried Out by the Engineering Department, National Physical Laboratory, England", M.A.P. Scientific and Technical Memoranda No. C1/42. (Summary). Controlled distribution.

8. L. L. Wyman: "Final Report on High Temperature Properties of Light Alloys", (NA-137): Part I - Aluminum. OSRD No. 3607, Serial No. M-251, April 15, 1944. Restricted.

9. Zay Jeffries: War Metallurgy Committee, Advisory Report on "The Effect of Impurities in Aluminum Alloys", (NA-119), to Office of Coordinator of Research and Development, Navy Department; OSRD No. 1731, Serial M-124, August 18, 1943. Restricted.

10. L. Rontgen and W. Koch: "Influence of Heavy Metals on Aluminum Alloys", Part II, Metallkunde, Vol. 26, pg. 9, 1934.

11. F. Bollenrath and H. Grober: "Anderung der Festigkeitseigenschaften Einiger Aluminiumlegierungen Durch Langzeitiges Erwarmen", Luftfahrtforschung, (Band 20,)Lfg. 10.

BIBLIOGRAPHY (Cont'd.)

12. N. Dudzinski and E. I. Brimelow: "New Aluminum-silicon Alloys of the Eutectic Type Suitable for Parts of Aero Engines Operating at Moderate Temperatures", Royal Aircraft Establishment, Farnborough, England; Report No. 8031, October, 1941. Controlled distribution.

13. N. Dudzinski and E. I. Brimelow: "Investigation of Silicon-Aluminum Eutectic Alloys of the Type Previously Examined by the Bristol Aeroplane Co., Royal Aircraft Establishment, Farnborough, England"; Report M. 7327, August, 1943. Controlled distribution.

14. J. K. Stanley: "Effect of Variables on the Recrystallization of Silicon Ferrite in Terms of Rates of Nucleation and Growth", Metals Technology, August, 1945. T.P. 1840.

15. J. J. Kanter: "Temperature Coefficient of Tensile Creep Rate", Transactions A.I.M.E., Iron and Steel Div., 1938.

16. J. Ciborski: N.R.L. Unpublished Report on "The Diffusion of Various Elements in Aluminum". 1946.

17. L. F. Mondolfo: Metallography of Aluminum Alloys, John Wiley & Sons, 1943

18. Ibid, pgs. 49-50.

19. W. D. Clark, J. Inst. Metals, 66, 271, 1940.

20. A.S.T.M. Standards, 1939, Part I; Specifications for Aluminum Sheet and Plate (B25-38T), p. 1015.

21. Ibid, High-Temperature Tension Tests of Metals (E21-37T), p. 1254.

22. S. Smirnov: "Secondary Aluminum Alloys. Improved Mechanical Properties and Finer Grain With Cerium Additions". Abstracted in Metals Industry, March 30, 1945.

Original data recorded in NRL Log Book No. 5394.

Table 1

Chemical Composition of Alloys

Heat or Alloy No.	Addition Element - Percent Cu	Si	Ni	Fe	Mg	Mn	Cr	Co	W	Mo.	V	Ti	B	Ce	Total Other rare Earths
						Binary Alloys									
1	2.81	0.066		0.049											
2	3.30	0.046		0.055											
3	4.85	0.045		0.053											
4	8.70	0.046		0.052											
5		0.25		0.046											
6		0.59		0.055											
7		1.41		0.060											
8		10.78		0.160											
9		0.085	0.94	0.062											
10		0.066	2.67	0.089											
11		0.055	5.88	0.136											
12		0.048		0.57											
13		0.056		1.03											
14		0.045		1.58											
15		0.043		4.98											
16		0.047		0.044	0.74										
17		0.055		0.039	4.97										
18		0.130		0.054	16.95										
18B		0.070		0.325	9.23										
54				0.03		1.44									
55				0.038		1.61									
56				0.026		1.89									
19		0.08		0.046		5.34									
58				0.058			0.63								
59				0.062			0.78								
60				0.062			0.88								
20		0.054		0.062			1.96								
21		0.074		0.077			4.90								
22		0.042		0.050				0.48							
23		0.044		0.054				1.05							
24		0.044		0.082				4.20							
25		0.041		0.140					0.53						
26		0.052		0.240					1.34						
27		0.067		0.500					4.48						
28		0.043		0.083						0.35					
29		0.042		0.170						0.76					
30		0.050		0.320						2.77					
31		0.045		0.063							0.15				
32		0.077		0.120							0.93				
33		0.040		0.143							2.76				
34		0.040		0.110								0.32			
M-8		0.043		0.69								3.85			

Table 1 (Continued)

Chemical Composition of Alloys

Heat or Alloy No.	Addition Element - Percent Cu	Si	Ni	Fe	Mg	Mn	Cr	Co	W	Mo	V	Ti	B	Ce	Total Other Rare Earth	Be
M-10		0.085		0.068									0.44			
35		0.050		0.180										3.31	3.05	
36		0.060		0.247										4.56	4.11	
Polynary Alloys																
37		1.00		0.045	4.77											
38		1.89		0.025	4.82											
39		1.51		0.130	4.93							0.12		0.20	0.37	
40		1.08		0.48	4.94	0.63										
41	0.60	1.10	0.43	0.64	4.76											
42		0.51		0.92	3.38	1.58		1.23								
Master Alloys																
M-1	49.3			0.06												
M-2		25.6		0.21												
M-3				0.05		19.3										
M-4			18.33	0.28												
M-5				0.13				9.44								
M-6				0.10			5.10									
M-7				13.20												
M-8		0.043		0.69								3.85				
M-9		0.08		0.13							2.10					
M-10		0.085		0.068									0.44			
M-11				0.55					4.83							
M-12				0.61						5.14						
M-13				1.25												2.75
M-14		0.10		0.63										14.85	15.25	

Table 2

Details of Melting Practice

Heat and Alloy No.	Addition of Master Alloy or Element	Addition Temp.	Max. Temp. of Heat	Degassing Temp.	Pouring Temp.	Permanent Mold Temp. (Pre-heat)	Pouring Time Sec.	Casting Quench Temp. c.w.q.	Remarks
1	Master Cu 49.3%	650°C	802°C	788°C.	749°C	104°C	6.5	454 to 482°C.	
2	Master Cu 49.3%	654	816	771 to 816	777	107	8.5	454 to 482°C.	
3	Master Cu 49.3%	704	785	782	777	119	6.8	454 to 482°C.	
4	Master Cu 49.3%	716	766	749 to 766	732	154	7.	454 to 482°C	Pouring Temp. Sli. Low
5	Master Si 25.6%	732	804	788 to 804	793	138	5.6	454 to 482°C	
6	Master Si 25.6%	721	838	799 to 788	793	149	6.	454 to 482°C	
7	Master Si 25.6%	760	782	771	771	141	5.	454 to 482°C	
8	Master Si 25.6%	754	827	738	746	116	6.	454 to 482°C	Pouring Temp.Sli. High
9	Master Ni 18.3%	732	827	785	788	188	10.	454 to 482°C	Slow pour, Pouring Temp.High
10	Master Ni 18.3%	732	804	782	760	157	5.2	454 to 482°C	
11	Master Ni 18.3%	738	816	760 to 816	749	149	5.4	454 to 482°C	Degass. Temp. Range Wide
12	Master Fe 13.2%	816	816	777	777	166	7.6	454 to 482°C	
13	Master Fe 13.2%	816	821	793	766	149	7.2	454 to 482°C	
14	Master Fe 13.2%	800	816	None	816	149	5.	454 to 482°C	Not Degassed-Pour.Temp.High
15	Master Fe 13.2%	832	832	804	771	168	5.	454 to 482°C	
16	Mg	682	804	799 to 777	760	149	4.8	454 to 482°C.	
17	Mg	680	843	771 to 788	777	127	5.5	454 to 482°C.	
18	Mg	682	777	771	716	143	7.0	454 to 482°C.	
18B	Mg	750	788	677 to 649	671	246	10.	393°C.	

Table 2 (Cont'd.)

Details of Melting Practice

Heat and Alloy No.	Addition Temp. of Master Alloy or Element Addition	Temp.	Max. Temp. of Heat	Degassing Temp.	Pouring Temp.	Permanent Mold Temp. (Pre-heat)	Pouring Time Sec.	Casting Quench Temp. c.w.q.	Remarks
19	Master Mn 19.3%	804°C	827°C	760 to 777°C	749°C	149°C	11.4	454 to 482°C.	Slow Pour
20	Master Cr 5.1%	816	816	771	760	190	9.9	454 to 482°C.	
21	Master Cr 5.1%	All Master Cold Start	788	788	788	190	9.9	454 to 482°C.	High dross, Stirred Well Before Pour
22	Master Co 9.4%	760	788	732	704	204	5.6	454 to 482°C.	Alloys 22-26 Stirred Well
23	Master Co 9.4%	771	738	721	716	182	5.3	454 to 482°C.	Metal dirty, Remelted.
24	Master Co 9.4%	746	771	760 to 704	732	196	5.2	454 to 482°C.	Preso-liqified
25	Master W 4.8%	746	738	732 to 704	704	170	5.3	454 to 482°C.	
26	Master W 4.8%	816	816	704 to 732	716	196	9.7	454 to 482°C.	
27*	Master W 4.8%	All Master Cold Start	843	821 to 843	838	232	4.1	454 to 482°C.	Metal Dirty-Remelted
28	Master Mo 5.14%	732	721	721 to 749	727	204	6.0	454 to 482°C.	
29	Master Mo 5.14%	760	771	743 to 716	721	238	4.0	454 to 482°C.	Metal Dirty-Remelted
30	Master Mo 5.14%	688	816	788 to 816	782	210	8.0	454 to 482°C.	Preso-lidified
31	Master V 2.1%	732	704	693 to 704	699	216	8.1	454 to 482°C.	
32	Master V 2.1%	All Master Cold Start	788	771 to 788	782	185	6.0	454 to 482°C.	Metal Dirty-Remelted
33*	Master V 2.1%	727	782	727	777	248	9.5	399	

Table 2 (Cont'd.)

Details of Melting Practice

Heat and Alloy No.	Addition of Master Alloy or Element Addition	Temp.	Max. Temp. of Heat	Degassing Temp.	Pouring Temp.	Permanent Mold Temp. (Pre-heat)	Pouring Time Sec.	Casting Quench Temp. c.w.q.	Remarks
34	Master Ti 3.85%	760°C.	788°C.	760 to 782°C.	782°C.	177°C	10.0	399°C.	
M-8	Powdered Ti	1149 to 1343	1200	760	760	180	4.0	470	Metal Dirty-Remelted Recovery Low
M-10	Powdered B	1149 to 1316	1200	760	760	238	5.0	470	Metal Dirty-Remelted Recovery Low
54	Electro Mn	900	750	Not Degassed	700	Approx. 150	Approx. 10.0	←--	These heats made in a different investigation Each heat 12 lbs.
55	Electro Mn	900	750	Not Degassed	700	Approx. 150	Approx. 10.0	---	
56	Electro Mn	900	750	Not Degassed	700	Approx. 150	Approx. 10.0	---	
58	Hi Purity Cr	900	750	Not Degassed	700	Approx. 150	Approx. 10.0	---	'
59	Hi Purity Cr	900	750	Not Degassed	700	Approx. 150	Approx. 10.0	---	'
60	Hi Purity Cr	900	750	Not Degassed	700	Approx. 150	Approx. 10.0	---	'
35	Master Ce 14.8% RE 15%	732	749	677	693	182	8.0	416	
36	Master Ce 14.8% RE 15%	704	771	710	704	160	6.0	377	
37	Si M$^{(x)}$ - Mg -	704 677	710	677 to 693	693	195	7.0	354	
38	Si M - Mg -	704 677	760	621 to 704	704	177	8.0	371	
39	Ti M - Si M - Ce M - Mg -	760 816 760 760	860	760 to 732	732	179	9.0	399	
40	Mn M Fe M > - Si M - Mg -	771 732 732	760	732	716	204	6.5	354	

Table 2 (Cont'd.)

Details of Melting Practice

Heat and Alloy No.	Addition Temp. of Master Alloy or Element Addition	Temp.	Max. Temp. of Heat	Degassing Temp.	Pouring Temp.	Permanent Mold Temp. (Pre-heat)	Pouring Time Sec.	Casting Quench Temp. c.w.q.	Remarks
41	Fe M Ni M > -	782°C.	816°C.	749 to 716°C	727°C	160°C.	10.0	382°C.	
	Si M Cu M > -	771							
	Mg -	749							
42	Co M Mn M > -	816	816	693 to 716	716	160	6.2	382	
	Fe M Si M > -	816							
	Mg -	738							

(x)-M = Master Alloy

Heats Degassed 4 minutes, dry N_2

* 4 lb. heat - all other heats 6 lbs.

RESULTS OF SHORT-TIME TENSILE TESTS

(Strip Specimens)

Alloy No.	%(1) Comp.		Tensile Strength-PSI				Yield Strength-PSI (0.2% offset)				% Elongation in 2 in.				Heat Treatment and Remarks
			Room Temp.	200°C	300°C	400°C	Room Temp.	200°C	300°C	400°C	Room Temp.	200°C	300°C	400°C	
			Each Value is the Ave. of at least 2 Determinations												
1	Cu	2.81	29620	22880	3370	1880	14400	12100*	2580	950	18	19	43	65	Note: For alloys 1-17 inc. speed of crosshead at all temps. was .05"/min up to (and thru) Y.P. and 0.1"/min. beyond YP to max. load and 0.3"/min. max. beyond ult. load to rupture.
2	Cu	3.30	36830	26100	4180	1860	14270	13000*	3280	1170	26	22	43.5	81	Cu Group: All specimens solution treated 4 hrs. at 571°C, cwq. For Rm. Temp. tests, specimens aged 3 hrs. at 177°C. For 200°C. tests, specimens stabilized 9 days at 210°C, A.C. For 300°C. tests, specimens stabilized 9½ da. at 310°C, cwq. For 400°C. tests, specimens stabilized 8 days at 410°C. cwq.
3	Cu	4.85	53820	29560	3870	2100	25810	15780*	3040	1500	23	19	61	87	
4	Cu	8.70	55250	18300	4270	1910	29270	11870*	2970	1260	17	21	57	81	
5	Si	0.25	10210	4470	1880	930	3330	2000	950	460	38	62	67.5	81	Si Group: For room temp. tests, specimens stress relieved 2 hrs. at 316°C. before machining, F.C., and 1 hr. at 316°C after machining, F. C. For 200°C tests, specimens stabilized 11 da. at 210°C. A.C. For 300°C tests, specimens stabilized 11½ days at 310°C. cwq. For 400°C tests, specimens stabilized 16½ da. at 410°C, cwq.
6	Si	0.59	10950	5400	2030	960	3620	4200	940	520	36	52	68.5	94	
7	Si	1.41	12080	5770	1980	900	3800	3590	1080	510	37	53	71	87	
8	Si	10.78	20920 12740(a)	8790	4180	1760	10200* 7320(a)	6310	3140	1190	27 30(a)	39	60.5	69	
9	Ni	0.94	12070 18390(a)	5360	2130	1140	3180 12000*(a)	3250	1180	730	35 20(a)	51	75	70	Ni Group: For room temp. tests, (a) specimens stress relieved 2 hrs. at 310°C before mach. For room temp. tests, other specimens also stress relieved 1 hr. at 310°C after mach. For 200°C tests, specimens stabilized 14 da. at 210°C, A.C. For 310°C tests, specimens stabilized 14 da. at 310°C, cwq. For 410°C tests, specimens stabilized 16½ da. at 410°C, cwq.
10	Ni	2.67	15360 26110(a)	9680	2970	1425	4260 23900*(a)	5980	2120	960	31 19(a)	23	67.5	55	
11	Ni	5.88	23140 16290(a)	15300	5400	2270	11830 10340*(a)	13900*	3600	1370	24 13(a)	17	43	48	
12	Fe	0.57	11520 17620(a)	8250	2600	1490	2910 10170*(a)	5850	1460	990	43 14(a)	15.5	56	56	Fe Group: Treatments same as for Ni group.
13	Fe	1.03	13380 18490(a)	7420	3210	1790	3770 10000*(a)	5970	2150	970	33 19(a)	28	53	52	
14	Fe	1.58	15220 20000(a)	9210	3400	1930	3840 9200*(a)	6320	2430	1330	34 14(a)	26	43	45	
15	Fe	4.98	18920 20040(a)	9450	3830	2060	4260 16840*(a)	5800	2970	1540	20 11(a)	40	46	44	
16	Mg	0.74	14510 44670(a)	14820	3940	1540	4260 28110*(a)	10440*	2720	830	27 14(a)	17	76	100	Mg Group: Treatments same as for Ni group.
17	Mg	4.97	39240	21820	9320	3920	15800	13100*	5720	3210	26	48	95	46	
54	Mn	1.44	12650	Not Tested At This Temp.	2930	1670	3490	-	1480	1150	39	-	60	52	Note: For the remaining alloys in this tabl. the speed of crosshead remained the same as above except the tests at 400°C were made at 0.15"/min. instead of 0.1"/min. from Y.P. to beyond ultimate load
55	Mn	1.61	12780		3120	1820	3150	-	1840x	1130	35	-	61	62	Mn Group: All specimens sol. treated 66 hrs. at 610°C, c.w.q. For room temp. tests, specs. stress relieved 2 hrs. at 316°C before machining and 1 hr. at 310°C after machining, F.C. For 300°C. tests, specimens stabilized 9 da. at 310°C, cwq. For 400°C. tests, specimens stabilized 6 da. at 410°C, cwq.
56	Mn	1.89	13030		3060	1820	3190	-	1940	1170	33	-	56	61	
19	Mn	5.34	15510	"	4150	2760	3290	-	2610	2170	21	-	30	45	
58	Cr	0.63	10550	"	2970	1860	3510	-	1740	1430	41	-	58x	47	Cr Group: Treatments same as for Mn group.
59	Cr	0.78	12440	"	3120	2290	3870	-	1760	1520	40	-	50	45	
60	Cr	0.88	10350	"	3330	1820	4140	-	2420	1490	36	-	54	48	
20	Cr	1.96	11700	"	4700	2370	3940	-	2600	1950	26	-	40	37	
21	Cr	4.90	15220	"	5490	2980	4330	-	3080	2580	14	-	22	23	

Table 3

RESULTS OF SHORT-TIME TENSILE TESTS (Strip Specimens)

Alloy No.	%(1) Comp.	Tensile Strength-PSI				Yield Strength-PSI (0.2% offset				% Elongation in 2 in.				Heat Treatment and Remarks
		Room Temp.	200°C	300°C	400°C	Room Temp.	200°C	300°C	400°C	Room Temp.	200°C	300°C	400°C	
22	Co 0.48	9000	Not Tested At This Temp.	2090	1180	2860	-	1170	620	34	-	76	68	**Co group:** For room temp. tests, specs. stress relieved 2 hrs. at 316°C before machining and 1 hr. at 316°C after machining. F.C. For 300°C tests, specs. stabilized 25 days at 310°C. c.w.q. For 400°C tests, specs. stabilized 5 days at 410°C, c.w.q.
23	Co 1.05	12120		3650	1260	2970	-	1960	930	40	-	60	50	
24	Co 4.20	13600		4890	1640	3210	-	3200[x]	1050	18	-	40	58	
25	W 0.53	8900	"	2920[x]	1340	3010	-	1790[x]	950	37	-	47	57	**W Group:** Treatments same as for Co group.
26	W 1.34	10140	"	3450	2480	3170	-	2390	1140[x]	39	-	53	49	
27	W 4.48	12960	"	4640	2240 1090(3)	3400	-	2790	1650	30	-	39	41	
28	Mo 0.35	12850	"	2100	1540	6400	-	1020[x]	980	27	-	60	71	**Mo Group:** Treatments same as above.
29	Mo 0.76	13710	"	3300	1600	7030*	-	1600	1160	19	-	49	60	
30	Mo 2.77	15160	"	3950	2210	7660*	-	2580	1710	19	-	41	36	
31	V 0.15	9100	"	3280	2040	3490	-	2350[x]	1510	43	-	44[x]	53	**V Group:** Treatments same as above.
32	V 0.93	11500	"	5020	2020	3890	-	3230[x]	1720	36	-	35	48	
33	V 2.76	13510	"	Not Tested	2390	6620	-	-	1770	33	-	-	35	
34	Ti 0.32	12640	"	"	1880	7440*	-	-	1630	29	-	-	39	Ti Group, B, and rare earth group: Treatments at respective temperatures same as for Co group.
M-8	Ti 3.85	14360	"	5030	2100	8160*	-	3500[x]	1300	20	-	32[x]	35	
M-10	B 0.44	9850	"	2700	1370	3410	-	1580	750	27	-	59[x]	74	
35	Ce 3.3 RE 3.05(2)	18510	"	Not Tested	2640	9060*	-	-	1720	25	-	-	57[x]	
36	Ce 4.56 Re4.11(2)	20960	"	"	2840	9800*	-	-	1930	20	-	-	52	
		Polynary Alloys												
37	Mg 4.77 Si 1.0	35000	"	"	2970	14980	-	-	2320	21	-	-	71	**Polynary Alloys:** Room Temp. Tests, specs. stress relieved 2½ hrs. at 316°C. F.C. For 400°C, stabilized 5 days at 410°C c.w.q.
38	Mg 4.82 Si 1.89	25970	"	"	2700	10100*	-	-	1880	18	-	-	87	
39	See Table	23390	"	"	2740	8180*	-	-	2200	19	-	-	70	
40	"	38940	"	"	3240	19150	-	-	2280	13	-	-	70	**Note:** For all specimens tested at elevated temp. a minimum of 20 minutes was taken to bring specimens to temp. & were held at temp. for a minimum of 20 min. before applying load.
41	"	38280	"	"	3720	18930	-	-	2690	11	-	-	98	
42	"	40870	"	"	4090	18800	-	-	2690	17	-	-	83	
Base Alloy	Al 99.95	7000	2400	1850	900	2800	1900	600	500	41	73	80	116	R.T. Values from BuStds data book. High temp. values extrapolated from Kennedy's work (x)

Table 3 (continued)

Footnotes for Table 3

(1) See Table 3 for complete composition.

(2) Total other rare earths - mainly lanthanum.

(3) Determined at 0.1"/min. crosshead speed.

(a) Approximately 1/4 to 1/2 H temper - not completely stress relieved.

(x) Questionable stress-strain curve - Y.S. taken as average of two values.

(x) Value questionable.

Table 3 (continued)

Table 4

Percent Decrease Between The Tensile Properties Obtained At 300° And 400°C. For 25 Selected Alloys

Alloy No. (1)	Percent Addition Element	Percent Decrease In Tensile Str.(1)	Order	Percent Decrease In Yield Str.	Order	% Change In Elong. in 2 in.	Relative Order at 400°C. of Tensile Str.	Relative Order at 400°C. of Yield Str.
59	Cr 0.78	26.6	1	13.6	1	10.0 D(2)	9	12
26	W 1.34	28	2	52.3	16	7.5 D	6	22
19	Mn 5.34	33.4	3	16.8	3	50.0	4	3
31	V 0.15	37.8	4	35.7	6	25.0	16	(7)
56	Mn 1.89	40.5	5	39.6	8	8.9	22	18
14	Re 1.58	43.2	6	45.2	12	4.6	18	15
30	Mo 2.77	44.	7	33.7	5	12.2 D	12	8
21	Cr 4.90	45.7	8	16.2	2	4.5	2	2
3	Cu 4.85	45.7	9	50.6	15	42.6	13	13
15	Fe 4.98	46.2	10	48.1	14	4.3 D	15	11
20	Cr 1.96	49.5	11	25.	4	7.5 D	8	4
27	W 4.48	51.7	12	40.8	9	5.1	11	9
10	Ni 2.67	52.	13	59.4	21	18.5 D	25	23
34	Ti 0.32	54.	14	41.8	10	6.8	20	10
4	Cu 8.70	55.2	15	57.5	20	42.1	19	19
2	Cu 3.30	55.5	16	64.3	24	86.2	21	20
35	Ce 3.30 RE 3.05	56.7	17	55.8	17	21.3	5	(7)
36	Ce 4.56 RE 4.11(3)	57.6	18	57.1	19	30.	3	5
8	Si 10.78	57.8	19	62.	(22)	14.	23	17
17	Mg 4.97	58.	20	43.8	11	57.8 D	1	1
11	Ni 5.88	58.	21	62.	(22)	11.6	10	14
M-8	Ti 3.85	58.2	22	62.8	23	9.3	14	16
32	V 0.93	59.7	23	46.7	13	37.1	17	(7)
33	V 2.76	60.1	24	56.8	18	2.9	7	6
24	Co 4.20	66.4	25	36.	7	45.	24	21

(1) In order of increasing percentage difference in Tensile Str. between 300 and 400°C.

(2) "D" means Elongation decreased in going from 300 to 400°C - all other values represent increases in elong. in 2 inches.

(3) Total other rare earths, mainly lanthanum.

Table 5

Results of Hardness Tests
Hardness (Rockwell E and H Scales)

Alloy	As Cast(1)		Before Tensile Testing At Room Temp.		Strip Specimens Taken at Room Temperature After Stabilizing And Tensile Testing At 200°C		300°C		400°C		Remarks(2)
	Brinnell 10/500/30	RE	RE	RH	RE	RH	RE	RH	RE	RH	
1	—	—	52	88	63	92	75	55	108*	55	Solution treated and artificially aged.
2	41	—	64	95	66	93	77	62	116*	58	"
3	—	—	88	105*	78	101*	80	65	110*	64	"
4	61	—	93	107*	86	104*	82	66	127*	68	"
5	—	—	*	13	*	16	—	15	—	15	Stress relieved
6	24	—	*	15	*	16	—	18	—	13	"
7	—	—	*	24	*	34	—	12	—	14	"
8	50	—	25	71	*	66	—	67	—	58	"
9	—	—	—	27	*	24	—	10	—	11	"
10	30	—	35	50	*	68	—	28	—	45	"
11	42	—	65	80	56	83	—	79	19	65	"
12	—	—	*	19	*	53	—	10	—	13	"
13	25	—	29	40	*	55	—	30	—	26	"
14	—	—	30	50	*	68	—	32	—	30	"
15	37	—	41	59	*	64	—	56	—	50	"
16	—	—	42	37	41	78	66	35	85	32	"
17	54	—	79	93	75	98	108*	93	58	82	"
18	—	—	—	—	—	—	—	—	—	—	—
18B	80	82	—	—	—	—	—	—	—	—	—
22	—	—	—	22	Not Tested at 200°C		—	10	—	11	Stress relieved
23	28	*	—	33			—	29	—	26	"
24	36	13	—	46			—	46	—	45	"
25	—	—	—	30			—	11	—	13	"
26	23	*	—	32			—	20	—	29	"
27	31	14	—	46			—	45	—	47	"
28	—	—	—	74			—	12	—	12	"
29	24	*	—	66			—	20	—	20	"
30	31	*	—	68			—	44	—	40	"
31	—	—	—	26			—	20	—	18	"
32	27	*	—	38			—	34	—	35	"
33	27	*	*	45			Not tested at 300°C		*	30	"
34	19	*	*	30					*	15	"
M-8	—	—	—	67			—	57	—	41	"
M-10	25	*	—	36			—	22	—	22	"

Table 5 (Cont'd.)

Alloy	As Cast(1)		Before Tensile Testing At Room Temp.		Strip Specimens Taken at Room Temperature After Stabilizing And Tensile Testing At 200°C		300°C		400°C		
	Brinnell 10/500/30	RE	RE	RH	RE	RH	RE	RH	RE	RH	Remarks(2)
54	---	---	88	55	Not Tested at 200°C		88	35	87	32	Solution treated and stress relieved.
55	---	---	86	58			91	39	87	35	"
56	---	---	90	50			92	40	87	40	"
19	42	53	118*	65			110*	57	105*	57	"
58	---	---	85	35			86	35	86	33	"
59	---	---	85	37			86	36	87	35	"
60	---	---	84	40			87	30	85	36	"
20	34	15	92	50			91	41	91	44	"
21	34	16	110*	61			106*	55	109*	60	"
35	34	18	*	56			Not tested at 300°C		*	37	Stress Relieved
36	40	35	18	61					20	57	"
37	57	66	50	88					53	83	"
38	---	---	36	80					23	70	"
39	59	68	38	80					37	80	"
40	69	76	65	92					63	93	"
41	77	81	75	98					68	95	"
42	77	80	78	95					68	95	"

(1) Test made on 1-3/4" x 1-3/4" x 1/2" specimens sawed from castings - rough polished.

(2) Refer to heat treatment given strip specimens "before testing at room temp."

* OFF SCALE

Table 6

The Amounts of Single Additions Of 14 Elements Required To Give Optimum Tensile Properties To Aluminum At 300 and 400°C. After Prolonged Stabilization At Test Temperature

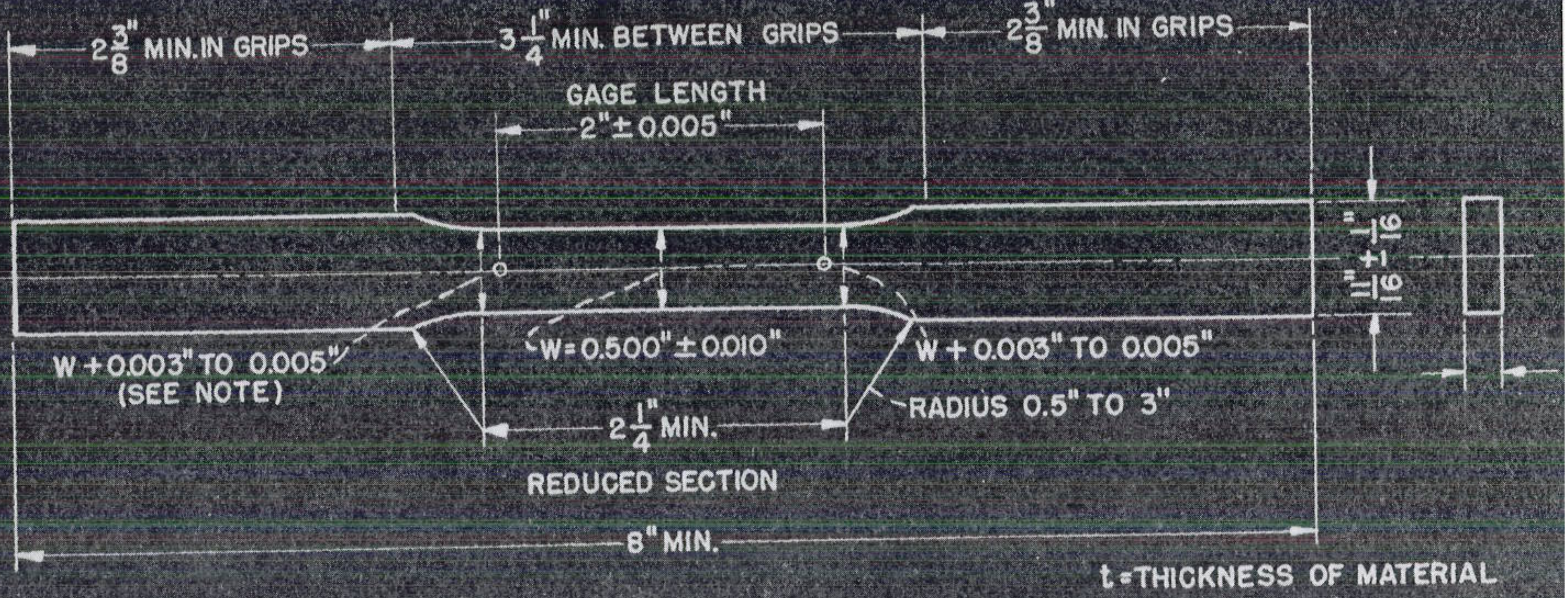
For wrought alloys only, tested at a cross head speed of 0.05 in./min. up to Y.P. and from 0.1 to 0.15 in./min. beyond Y.P. to ult. load.

Element*	% Required At 300°C	Element*	% Required At 400°C
Mg	4 - 6	Mg	4 - 6
Ce(1)	7 -14?	Cr	2 - 4
V	1 - 2	Ce(1)	7 -14?
Ni	4 - 6	Mn	5 - 6 (3.5-5 more practical)
Cr	2 - 3	W	1.5 - 2
Co	3 - 4	Ni	6 - 7 (4-5 more practical)
Ti	0.5- 2	V	0.15- 0.6
W	2 - 4	Mo	1.5 - 2.5
Cu	3 - 3.5	Cu	4.5 - 5.5
Si	10 -13	Fe	1.5 - 2
Mn	3.5- 5	Ti	0.4 - 0.8
Mo	1 - 2	Si	10 -13
Fe	1.5- 2	Co	2 - 2.5
B	0.2- 0.5	B	0.2 - 0.5

* Elements arranged in order of decreasing effectiveness at upper limits of the prescribed composition ranges.

(1) Not tested at 300°C, position estimated. Additions over 8% would probably have put Ce in second place at 400°C. Ce content actually refers to "Mischmetal".

STANDARD A.S.T.M. TENSION TEST SPECIMEN FOR SHEET METALS.



NOTE: GRADUAL TAPER FROM ENDS OF REDUCED SECTION TO MIDDLE.

ALL MACHINING DIMENSIONS ARE SHOWN BELOW AND TESTING DIMENSIONS ABOVE SPECIMEN.

THE ENDS SHALL BE SYMMETRICAL WITHIN 0.01 IN.

M-2855

LOADING BOLTS FOR TESTING ROUND AND FLAT SPECIMENS AT ELEVATED TEMPERATURES

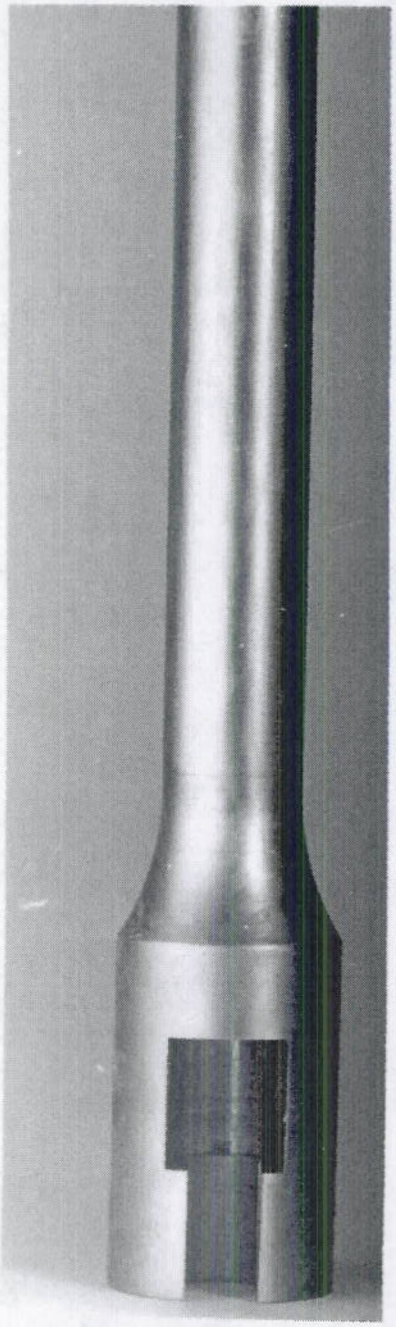


FIG. 1
(PARTIAL VIEW, 1X)

FIG. 2
(PARTIAL VIEW, 1X)

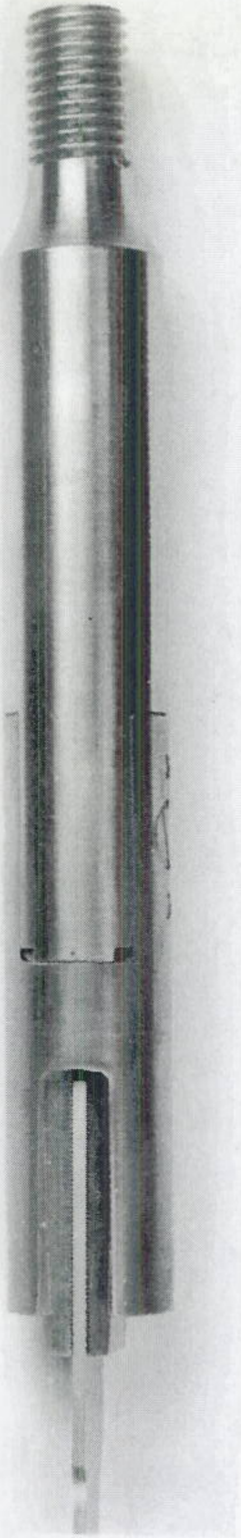
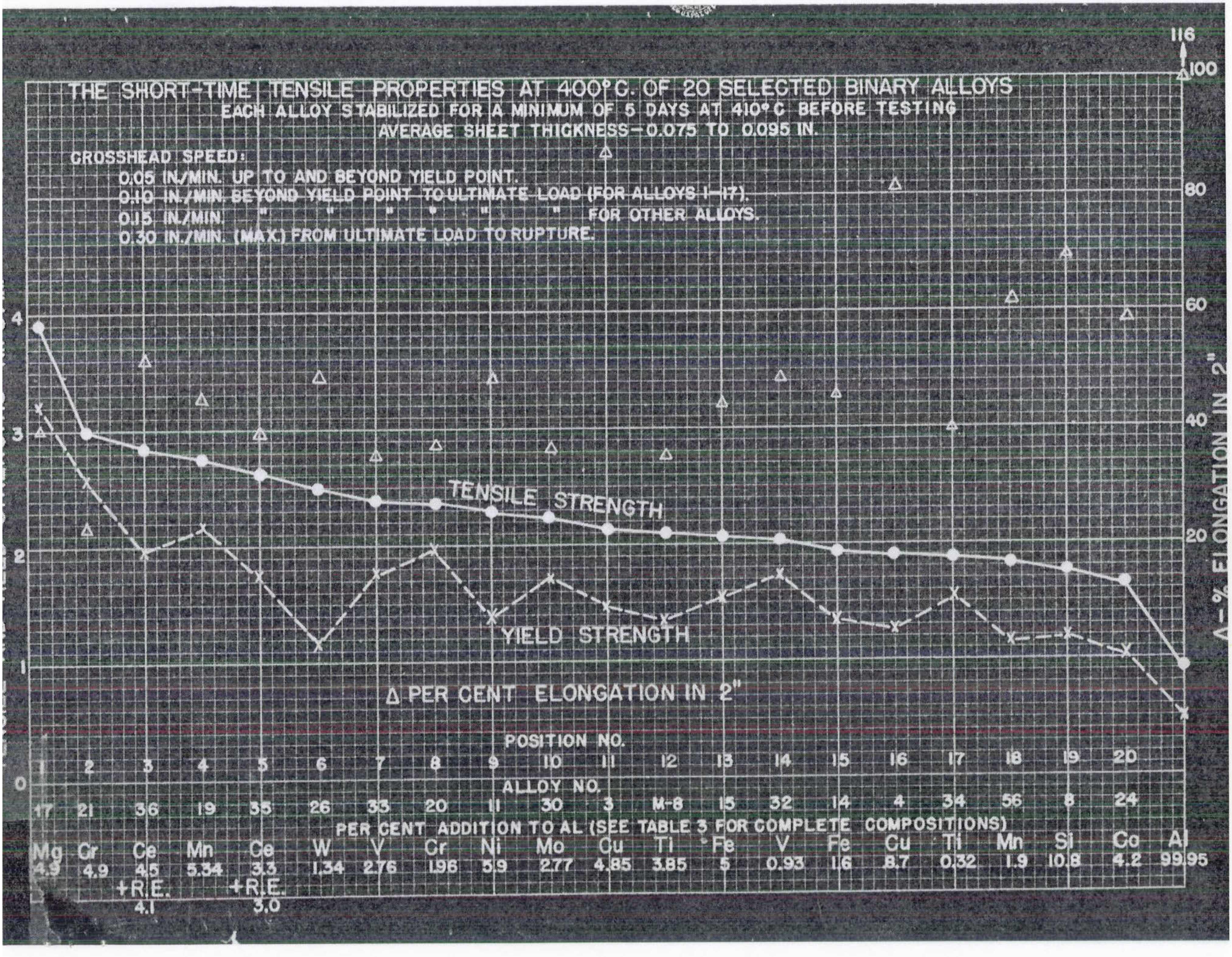


FIG. 3
(3/4X)



THE SHORT-TIME TENSILE PROPERTIES AT 400°C. OF 20 SELECTED BINARY ALLOYS

EACH ALLOY STABILIZED FOR A MINIMUM OF 5 DAYS AT 410°C BEFORE TESTING

AVERAGE SHEET THICKNESS–0.075 TO 0.095 IN.

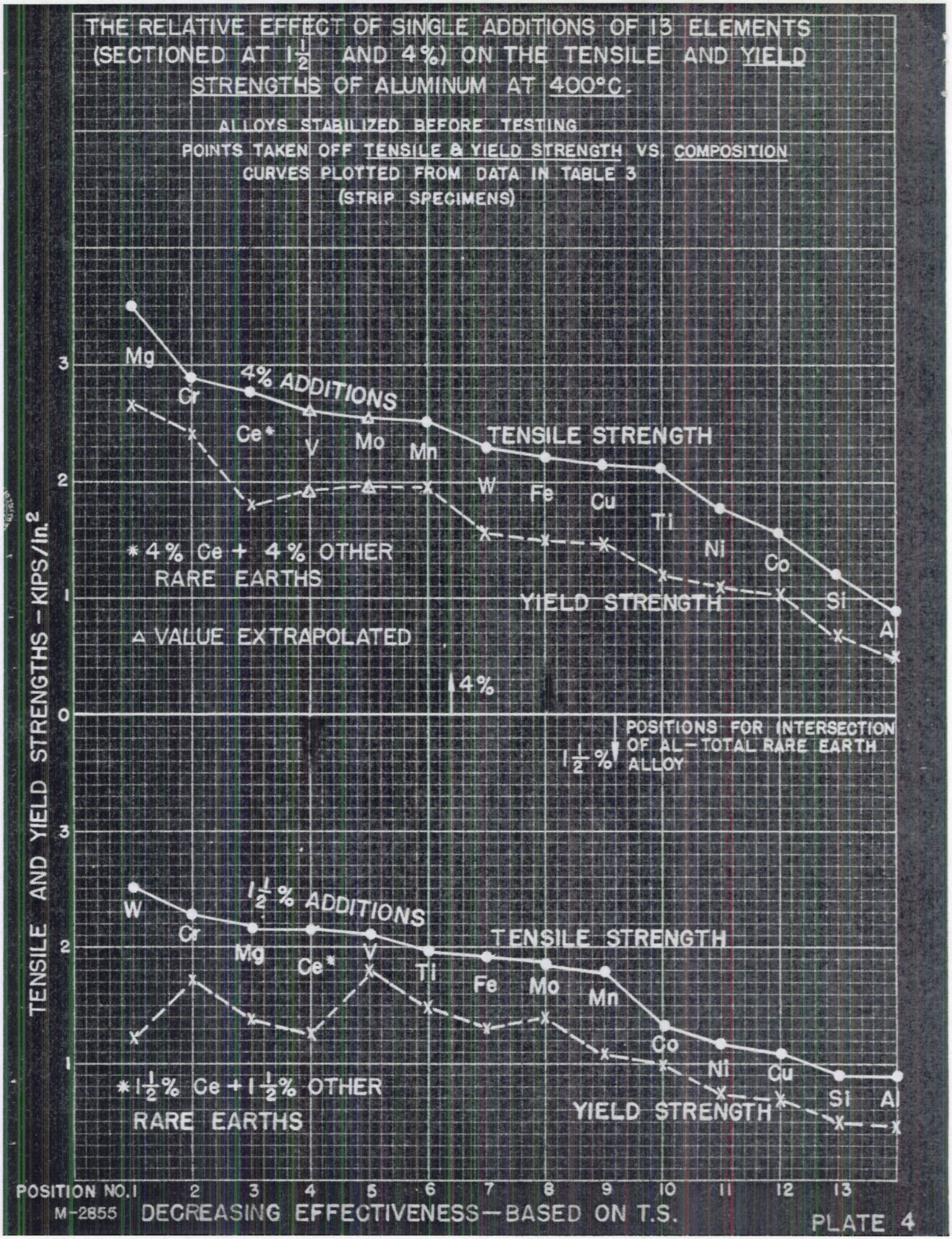
CROSSHEAD SPEED:

0.05 IN./MIN. UP TO AND BEYOND YIELD POINT.

0.10 IN./MIN. BEYOND YIELD POINT TO ULTIMATE LOAD (FOR ALLOYS 1–17).

0.15 IN./MIN. " " " " " FOR OTHER ALLOYS.

0.30 IN./MIN. (MAX.) FROM ULTIMATE LOAD TO RUPTURE.

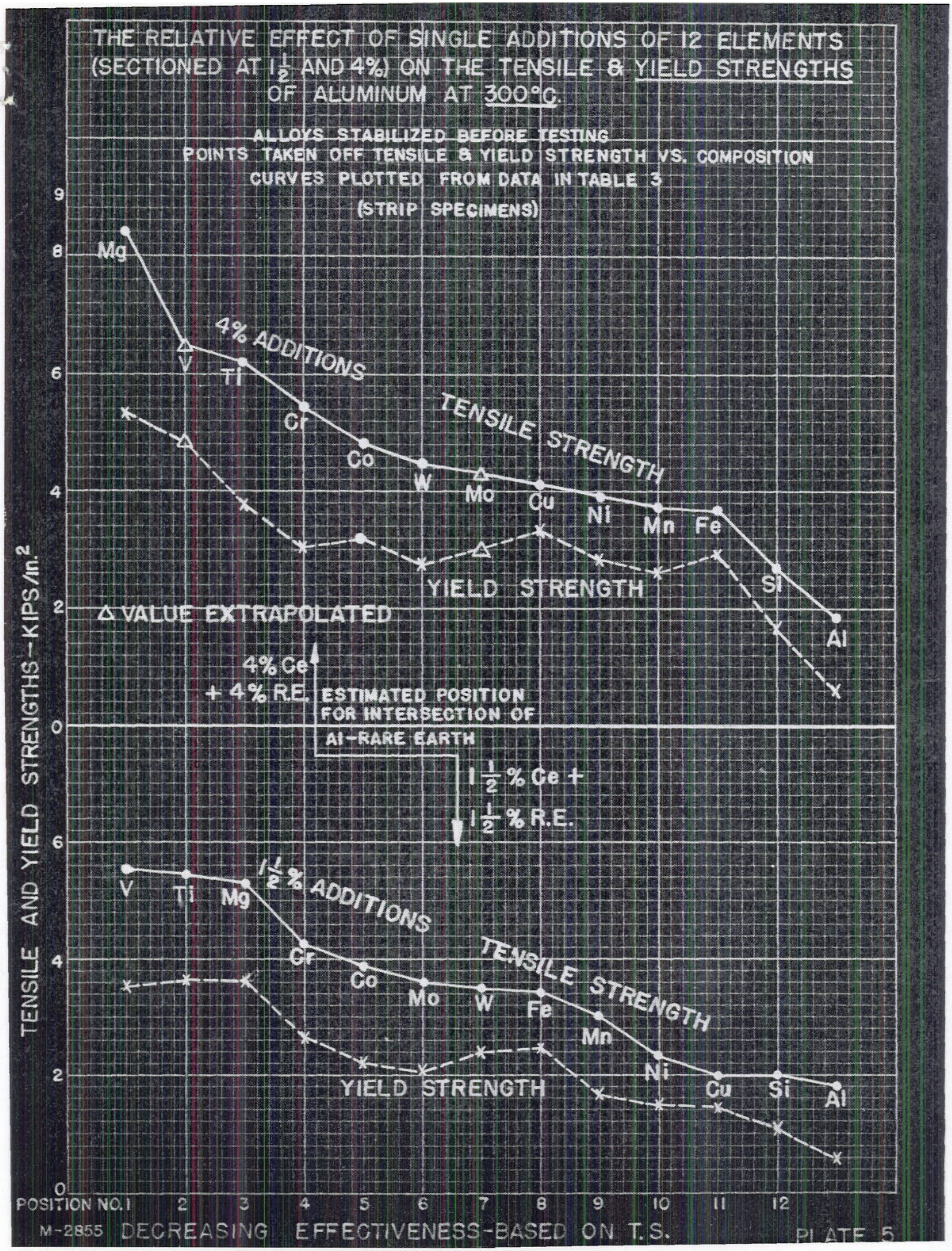


THE RELATIVE EFFECT OF SINGLE ADDITIONS OF 13 ELEMENTS (SECTIONED AT $1\frac{1}{2}$ AND 4%) ON THE TENSILE AND YIELD STRENGTHS OF ALUMINUM AT 400°C.

ALLOYS STABILIZED BEFORE TESTING
POINTS TAKEN OFF TENSILE & YIELD STRENGTH VS. COMPOSITION
CURVES PLOTTED FROM DATA IN TABLE 3
(STRIP SPECIMENS)

M-2855

PLATE 4



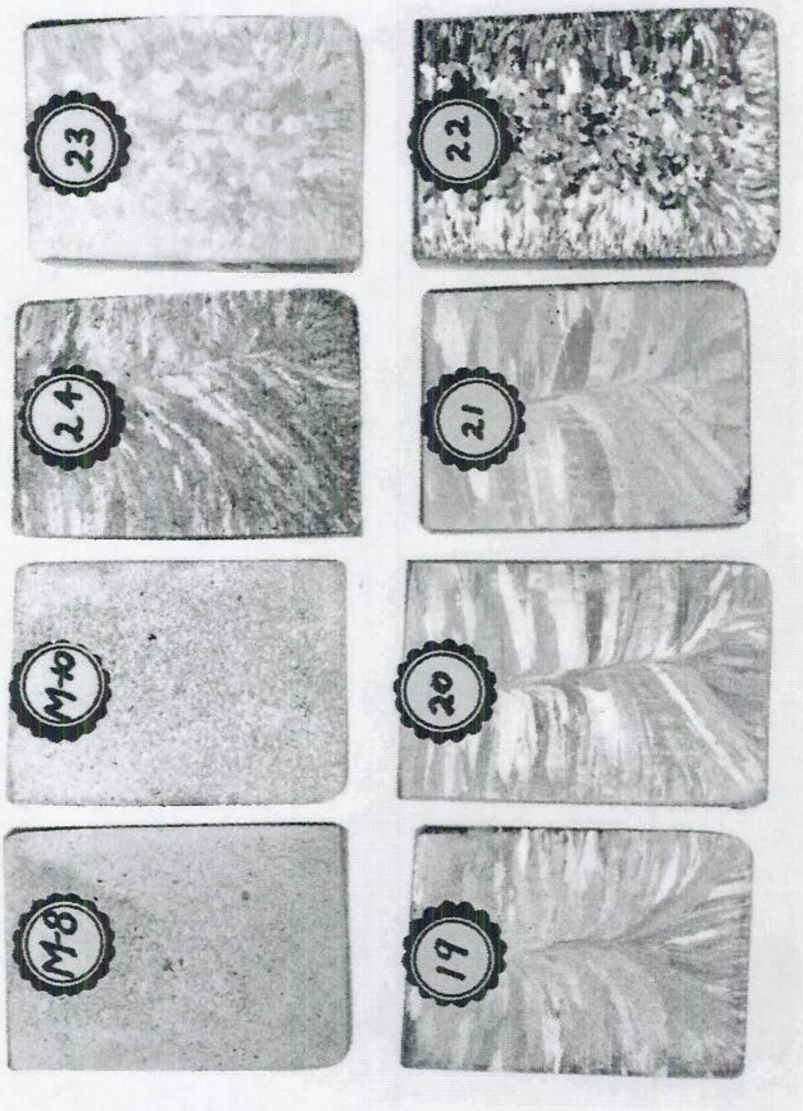
THE RELATIVE EFFECT OF SINGLE ADDITIONS OF 12 ELEMENTS (SECTIONED AT $1\frac{1}{2}$ AND 4%) ON THE TENSILE & YIELD STRENGTHS OF ALUMINUM AT 300°C.

ALLOYS STABILIZED BEFORE TESTING
POINTS TAKEN OFF TENSILE & YIELD STRENGTH VS. COMPOSITION CURVES PLOTTED FROM DATA IN TABLE 3
(STRIP SPECIMENS)

M-2855

PLATE 5

MACROSTRUCTURES OF PERMANENT MOLD CAST BINARY ALUMINUM ALLOYS



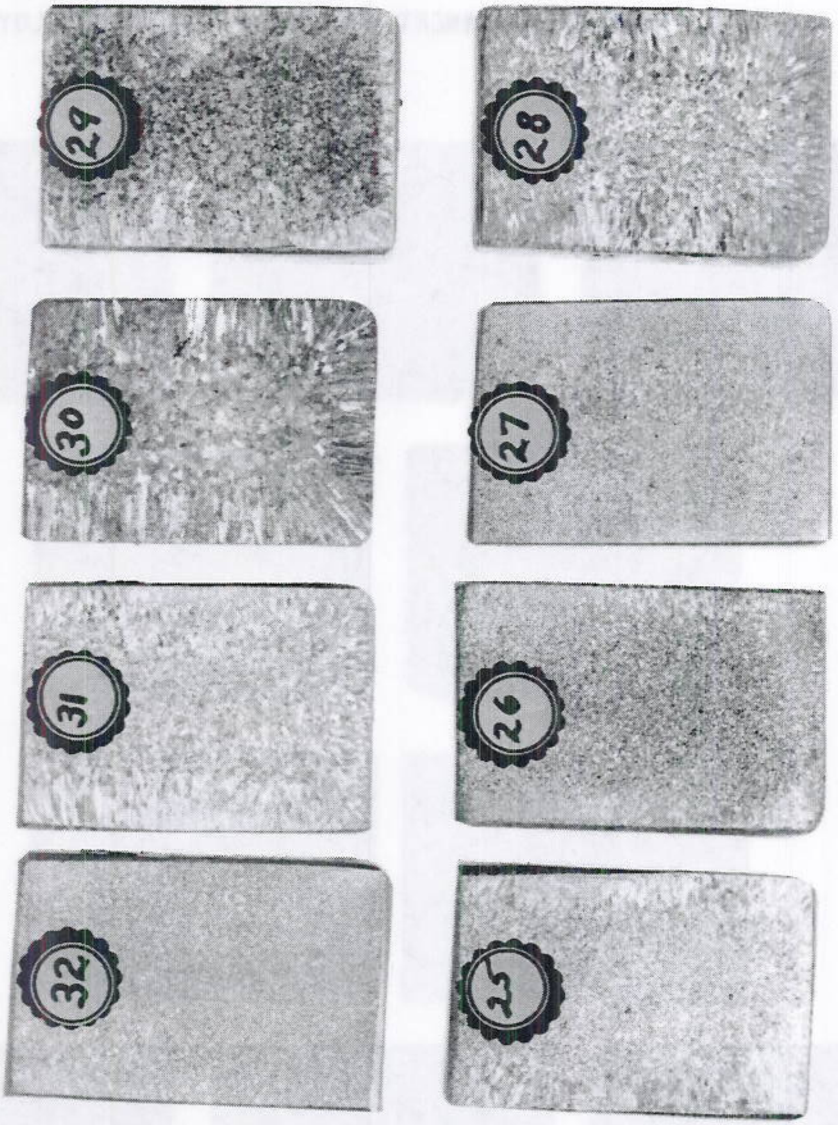
ALLOY	PER CENT COMPOSITION
19	Mn 5.34
20	Cr 1.96
21	Cr 4.90
22	Co 0.48
23	Co 1.05
24	Co 4.20
M-8	Ti 3.85
M-10	B 0.44

TUCKER'S ETCH

APPROXIMATELY 1X

PLATE 7

MACROSTRUCTURES OF PERMANENT MOLD CAST BINARY ALUMINUM ALLOYS



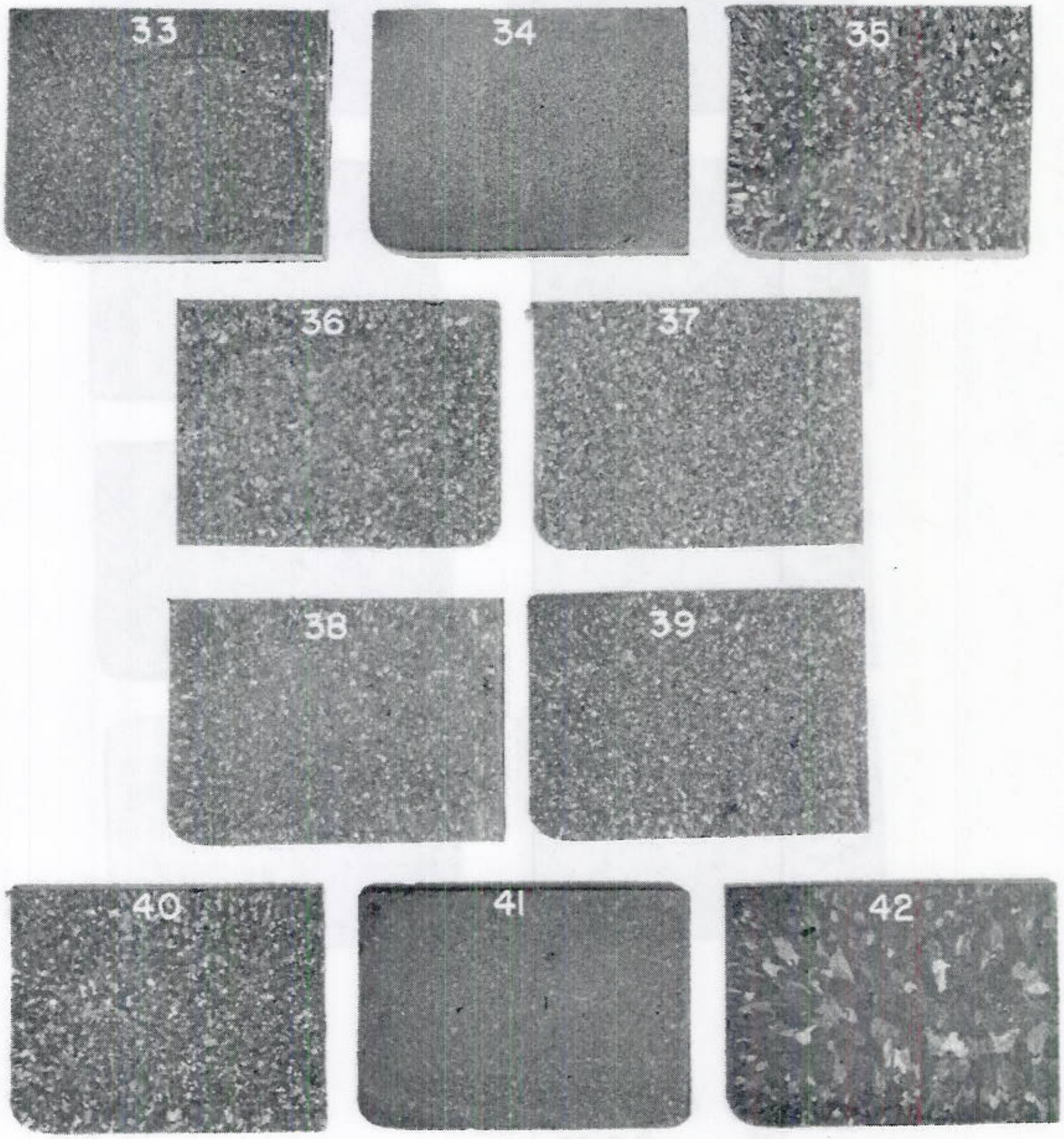
ALLOY	PER CENT COMPOSITION
25	W 0.53
26	W 1.34
27	W 4.48
28	Mo 0.35
29	Mo 0.76
30	Mo 2.77
31	V 0.15
32	V 0.93

TUCKER'S ETCH

APPROXIMATELY 1X

PLATE 8

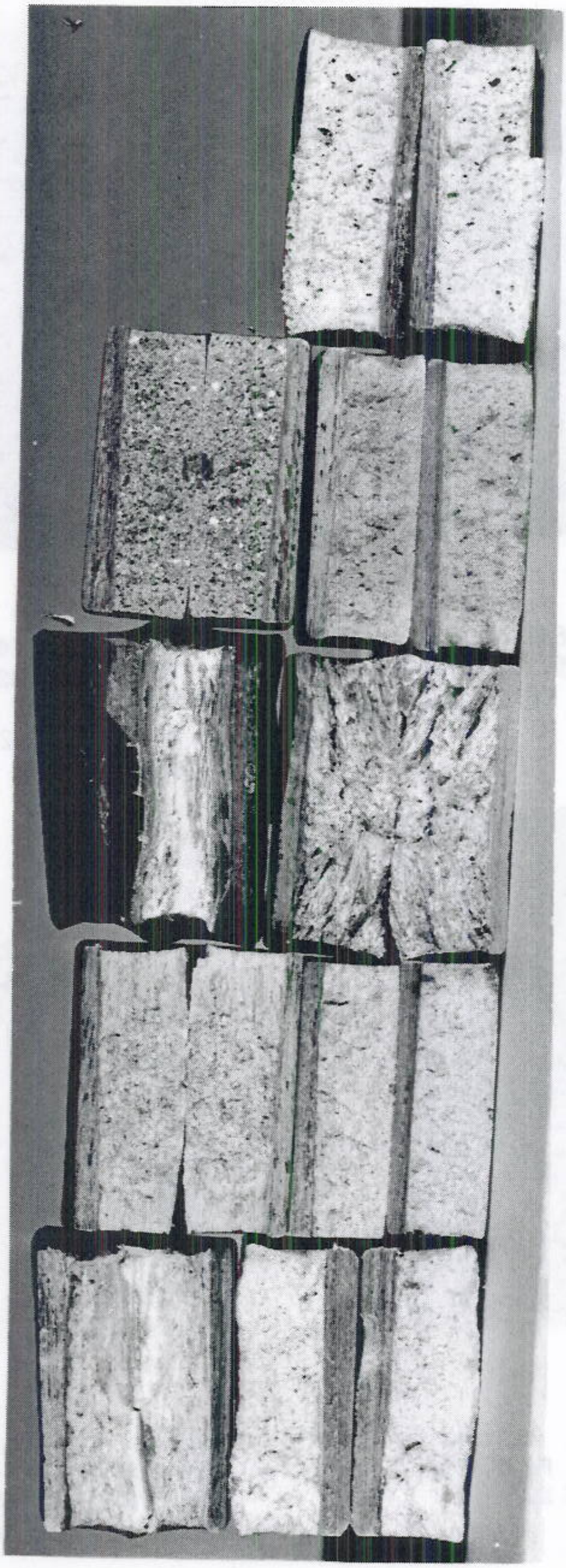
MACROSTRUCTURES OF FOUR BINARY, TWO TERNARY AND FOUR POLYNARY PERMANENT MOLD CAST ALUMINUM ALLOYS



33 - 36 BINARY   37 - 38 TERNARY   39 - 42 POLYNARY

SEE TABLE 3 FOR COMPOSITIONS
TUCKER'S ETCH
APPROXIMATELY 1X

TYPICAL FRACTURES OF PERMANENT MOLD CAST BINARY ALUMINUM ALLOYS



APPROXIMATELY 1X

PLATE 10

PHOTOMICROGRAPHS OF SELECTED ALLOYS

FIG. 1

NEEDLES OF B_2Al. PERMANENT MOLD CAST - B 0.44%, Fe 0.068%, Si 0.085%, 250X, 10% NaOH ETCH WITH CHROMIC ACID RINSE.

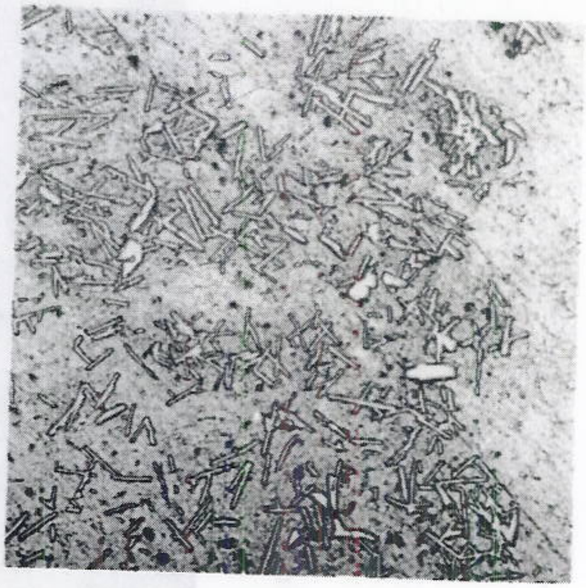


FIG. 2

PRIMARY NEEDLES OF ϵ WAl_{12} LIGHT, OXIDE INCLUSIONS DARK. PERMANENT MOLD CAST - W 5.1%, Fe 0.5%, Si 0.068%, 250X, ETCH SAME AS IN FIG. 1.

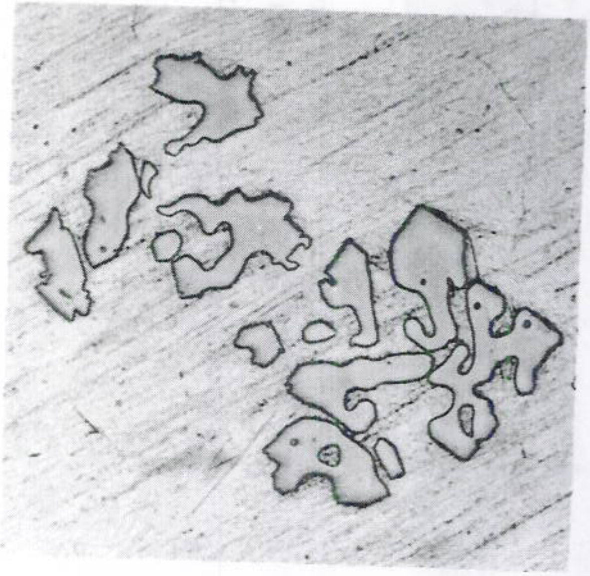


FIG. 3

PRIMARY CRYSTALS OF VAl_7. CAST IN PERMANENT MOLD - V 0.93%, Fe 0.12%, Si 0.077%. UNETCHED, 250 X.

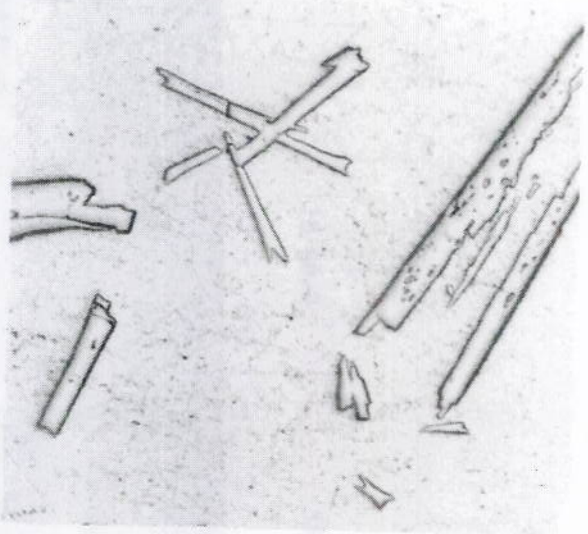


FIG. 4

PRIMARY CRYSTALS OF $MoAl_5$. CAST IN PERMANENT MOLD - Mo 0.76%, Fe 0.17%, Si 0.042%. UNETCHED, 250 X.

PHOTOMICROGRAPHS OF SELECTED ALLOYS

FIG. 1

PRIMARY CRYSTALS OF $TiAl_3$. CAST IN PERMANENT MOLD - Ti 3.85%, Fe 0.69%, Si 0.04%. UNETCHED, 250 X.

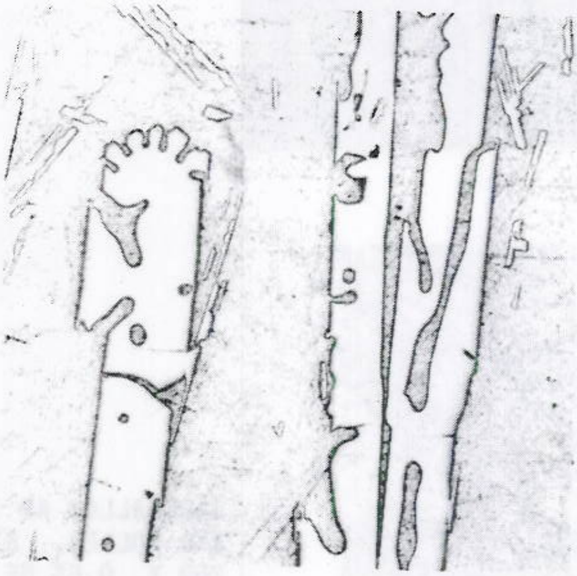


FIG. 2

PRIMARY CRYSTALS OF Co_2Al_9. CAST IN PERMANENT MOLD - Co 4.2%, Fe 0.08%, Si 0.04%, 250 X, 10% NaOH ETCH WITH CHROMIC ACID RINSE.

PHOTOMICROGRAPHS OF SELECTED ALLOYS

FIG. 1

PRIMARY CRYSTALS OF Fe_2Al_7 IN EUTECTIC Al-Fe_2Al_7. CAST IN PERMANENT MOLD - Fe 5%, Si 0.04%, 250 X, 0.5% HF ETCH.

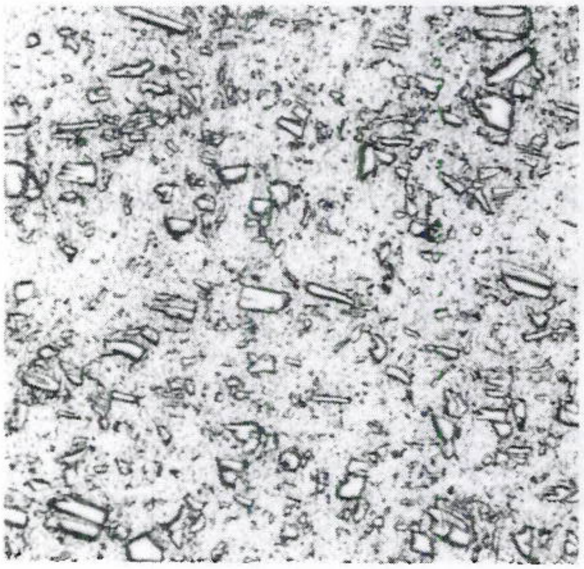


FIG. 2

SAME ALLOY AS IN FIG. 1, AS FORGED AND ROLLED. STABILIZED AT 410°C. 250 X, 0.5% HF ETCH.

PLATE 13

MACROSTRUCTURES OF THE WROUGHT Al-5Mg AND Al-6Ni ALLOYS BEFORE AND AFTER STABILIZING AND TESTING AT 400°C.

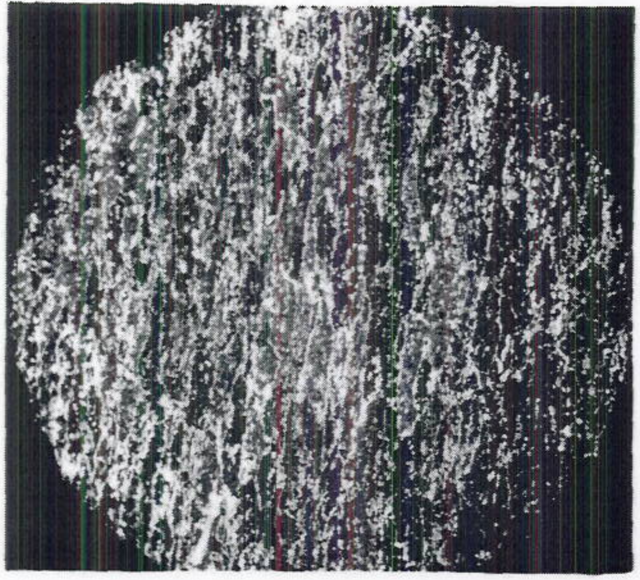


FIG. 1

STRIP FORGED AND ROLLED FROM PERMANENT MOLD CAST ALLOY - Mg 5%, Si 0.05%, Fe 0.04%. STRESS RELIEVED. 25X, TUCKER'S ETCH.

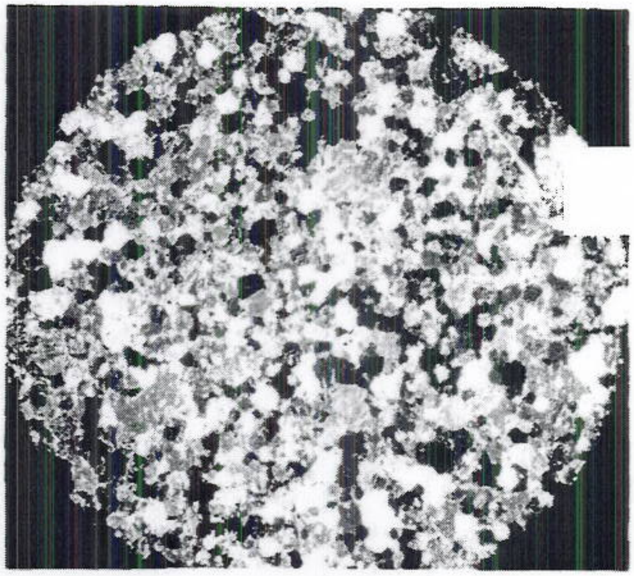


FIG. 2

SAME ALLOY AND FORMING TREATMENTS AS IN FIG. 1. STABILIZED 16 DAYS AT 410°C. - RECRYSTALLIZED. 25X, TUCKER'S ETCH.

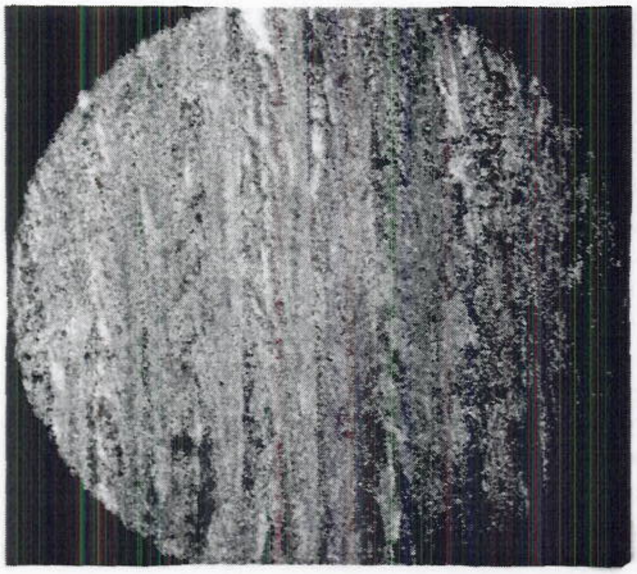


FIG. 3

FORMING TREATMENT SAME AS IN FIG. 1 Ni 5.88%, Fe 0.13%, Si 0.05%. STRESS RELIEVED. 25 X, TUCKER'S ETCH.

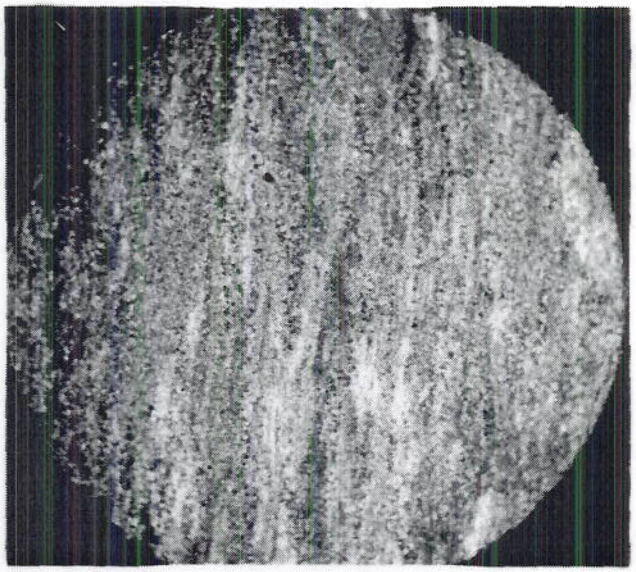


FIG. 4

SAME ALLOY AND FORMING TREATMENTS AS IN FIG. 3. STABILIZED 16 DAYS AT 410°C. 25 X, TUCKER'S ETCH.

(DIFFERENT FIELDS IN EACH CASE)

PLATE 14