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DEVELOPMENT OF HIGH TEMPERATURE NICKEL-BASE ALLOYS FOR JET ENGINE TURBINE BUCKET APPLICATIONS

By

S. T. SCHEIRER and R. J. QUIGG

Prepared for

NATIONAL AERONAUTICS AND SPACE ADMINISTRATION

Contract NAS 3-7267

TRW EQUIPMENT LABORATORIES

A DIVISION OF TRW INC. • CLEVELAND, OHIO 44117

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SEMI-ANNUAL REPORT

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October 20, 1965

Contract NAS3-7267

Technical Management
NASA LEWIS RESEARCH CENTER
Cleveland, Ohio
Spacecraft Technology Division

F. H. Harf, Project Manager
R. L. Dreshfield, Research Advisor

Materials and Processes Department
TRW Equipment Laboratories
TRW Inc.
23555 Euclid Avenue
Cleveland, Ohio 44117

FOREWORD

This First Semi-Annual Engineering Report covers the work performed under NASA Contract NAS3-7267 during the period from 1 April 1965 to 1 October 1965. The report has been given the NASA number CR-54504. The TRW internal number ER-6666 has been assigned.

This contract was initiated between NASA Lewis Research Center and TRW Inc. for the "Development of High Temperature Nickel-Base Alloys by Conventional Alloying Techniques for Application as Jet Engine Turbine Buckets". Technical direction is being supplied by Mr. F. H. Harf, Project Manager, of the Lewis Research Center, Spacecraft Technology Division, Cleveland, Ohio. Mr. R. L. Dreshfield is serving in the capacity of NASA Research Advisor.

Mr. S. T. Scheirer of TRW Equipment Laboratories is the project engineer in charge. Dr. R. J. Quigg, Research Supervisor, Superalloy Group, is the TRW Program Manager. Technical assistance was provided by Dr. H. E. Collins of the Superalloy Group. Messrs. J. T. Laughlin and C. L. Campbell assisted in the execution of the work.

TRW PUBLICATION REVIEW

Prepared by: S. T. Scheirer

S. T. Scheirer
Metallurgist

Approved by: R. J. Quigg

R. J. Quigg
Research Supervisor

Approved by: A. S. Nemy

A. S. Nemy
Manager
Materials and Processes Department

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I SUMMARY

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A program has been initiated under NASA sponsorship to develop a material with superior properties at elevated temperatures for utilization in turbine blade applications. It is proposed that the necessary high temperature strength can be provided in a nickel-base superalloy by utilizing the maximum capability of the three available strengthening mechanisms, intermetallic gamma prime precipitation (Ni_3Al), solid solution strengthening with refractory and precious metals, and stable carbide formations through the addition of strong carbide forming elements.

A stress rupture test at 2000°F and 15,000 psi has been formulated to approximate the target properties desired. Through the addition of varying amounts of refractory metals (Mo, W, and Ta) it has been possible to statistically analyze the effects of each in a basic superalloy composition containing fixed amounts of Co, Cr, C, B, Zr, and Ni at three separate levels of Al and Ti. Tantalum was found to be more beneficial to 2000°F stress rupture properties than Mo or W, with tungsten the second most effective. A total amount of refractory elements near 10 percent appeared effective if Al and Ti were kept relatively high. Greater amounts of refractory metal additions were advantageous if the aluminum and titanium were fixed at a lower level. Metallographic analysis correlated with the mechanical properties of the alloys; those having few strengthening phases were weak and ductile, while those having excessive amounts of intermetallic phases present in undesirable morphologies were brittle.

The best alloys developed thus far in this program compare favorably with the best commercially available alloys and possess reasonable ductility. The base alloy composition selected for further study contained 8.0%Ta, 5.5%W, and 2.0%Mo. The titanium level has been fixed at 1.0 percent, while the aluminum will be varied on a statistical basis between 4.5 percent and 6.3 percent. The next series of alloys will investigate additions of Cb, V, Hf, Ru, and Re.

Author

II INTRODUCTION

The quest for larger and faster aircraft has led to a continuing demand for more sophisticated power plants. As is often the case, limitations of the available materials retard progress from drawing board to hardware. Specifically, gas turbine engine efficiency is dependent upon operating temperature, which means that improved performance is contingent upon the temperature capability of materials in the hot section of the engine.

One such hot zone in the turbine engine, which is also subjected to considerable stress, is the rotating turbine itself. Currently fabricated from nickel-base superalloys, the turbine blades must be creep resistant for extended periods of time while operating at temperatures well above 1700°F. Present nickel-base superalloys have far surpassed the capabilities of the earliest alloys of the type. However, for increased engine efficiency, still higher temperature potential is required in a blading material. The additional necessities of oxidation and thermal fatigue resistance make the desired requirements still more stringent.

There exists, of course, more than one potential solution to the problem of presenting the designer with a more serviceable blading material for elevated temperature usage. Considering that today, the materials commonly used for turbine blades are coated superalloys of either the nickel-cobalt-, or iron-base systems, the first avenue of approach would be an improvement or upgrading in high temperature strength of one or more of these basic systems. A second approach is the use of present-day materials with elaborate cooling passages provided to insure that while the engine operates at higher ranges, the blades themselves are not subjected to unbearably high temperatures[1]. Another alternative solution is the use of refractory metals and alloys, together with an oxidation resistant coating[2]. Other possibilities perhaps exist. While attractive arguments can be formulated for any one of the previously mentioned solutions, the one most in keeping with past technological advances is the first, that of upgrading the properties of an existent superalloy system.

Consideration of the program target of high strength and 1875°F service eliminates both iron- and cobalt-base superalloys from further evaluation. Iron-base superalloys are presently not operable above 1500°F[3] where loss of strength and severe oxidation becomes prohibitive. Cobalt alloys, while possessing the capability to retain strength above 2000°F, currently cannot be strengthened to anywhere near the desired load bearing capacity[4]. Thus, it appears that the nickel-base superalloy system is the most likely candidate for improvement to meet the required properties. The present-day nickel-base superalloys and their pertinent properties will next be discussed.

A. Status of Nickel-Base Superalloys

Chronologically, nickel-base alloys have undergone a progressive development in the type of alloying constituents added. This has resulted in a steady improvement in high temperature strength over the original 20 Cr Nimonic

75 alloy which can be considered the forerunner of the present superalloys. Recently, development has been proceeding along two somewhat divorced routes, wrought alloys and cast alloys. Basically, the physical metallurgy involved varies only in degree between the two cases and a single treatment of the concepts involved will suffice for both. However, there are certain aspects of wrought and cast alloys which bear separate mention.

1. Physical Metallurgy

Three basic alloy strengthening mechanisms manifest themselves in nickel-base alloy systems: intermetallic formation, solid solution strengthening, and carbide strengthening. While certain alloying elements can contribute to more than one mechanism within a given alloy, the three are best discussed separately.

a. Intermetallic Formation

It is commonly held that the most important single factor contributing to the retention of high temperature strength in nickel-base alloys is the precipitation of the Ni_3Al intermetallic phase. In this compound, a substantial portion of the aluminum atoms can be replaced by titanium. Thus, a proper balance of both aluminum and titanium levels are necessary if this mechanism is to be an effective strengthening agent. Other elements, added for various reasons, may also have subordinate effects upon the Ni_3Al (gamma prime) phase. In particular, the heavy metals molybdenum and tungsten retard precipitate growth by reducing diffusion rates, and cobalt is reputed to enhance the high temperature stability of gamma prime by increasing the solutioning temperature of Ni_3Al.[5]

A variety of morphological forms may be associated with Ni_3Al depending both upon its stoichiometry and the thermal and mechanical history of a given specimen. One form, prevalent in lightly alloyed systems, is a fine intragranular precipitate which is developed by use of a solutioning and aging treatment, Figure 1A. In more heavily alloyed superalloys, it is often impossible to suppress the precipitation of gamma prime after solutioning, Figure 1B. Cast alloys, often have a gamma prime dispersion, which is favorable to high temperature properties, formed directly upon solidification and cooling. Such alloys, generally containing high percentages of aluminum and titanium, exhibit a coarse, somewhat blocky type of intragranular gamma prime, Figure 2A. Heat treated alloys containing larger amounts of addition elements also exhibit a blocky intragranular gamma prime morphology, Figure 2B, the relative coarseness of the particles depending both upon chemistry and heat treating temperature. These forms of gamma prime are considered to be beneficial to high temperature strength and are encouraged through the use of controlled casting techniques and selected heat treating procedures.

Another gamma prime morphology often observed in heavily alloyed cast superalloys is the blossom-like formation of gamma, the nickel-rich solid solution, and gamma prime, Figure 3. Formed directly upon solidification, this structure can be a detriment to high temperature properties. Careful control of casting procedures in an alloy of proper composition can usually minimize this effect.

b. Solid Solution Strengthening

Reference to binary phase diagrams[6] will reveal that a number of elements have a wide solubility range in nickel. Noteworthy among these are the refractory metal elements and cobalt. Recent trends in superalloy development have shown that increased amounts of refractory metal solid solution hardeners significantly improve the stress rupture properties, especially at 1800°F and above. Unfortunately, while increasing strength, some of the refractory metals possibly decrease oxidation resistance, ductility, and, particularly in the case of tungsten, increase the density of the alloy, an important consideration for aerospace applications.

Other elements which are taken into the nickel-rich gamma solid solution are chromium and cobalt. While not particularly beneficial to high temperature strength, these elements nevertheless do have important effects: chromium on oxidation resistance and carbide formation, and cobalt reportedly upon gamma prime stability (as previously mentioned) and workability and ductility.

c. Carbide Strengthening

Despite the fact that the carbon level in most nickel-base superalloys is quite low (approximately 0.10 per cent by weight), the formation of a variety of carbides is an important contribution to the strengthening of these metals. Nickel itself is not a strong carbide forming element, but the majority of the alloying constituents have a fairly high propensity to form one or more carbide types. Seemingly minor alterations in chemistry have a large influence upon the type, stability, and morphology of the carbides formed.

The most stable carbide formed in nickel-base superalloys is the MC type. The metallic component is generally titanium, although in certain cases it can also contain varying amounts of columbium. MC is generally a stable carbide close to the fusion temperature and assumes a massive somewhat cubic structure, Figure 4A, evenly distributed throughout the microstructure. However, in certain instances, such as in the alloy Waspaloy, MC is formed in grain boundary areas, Figure 4B, and is relatively unstable at temperatures well below the melting point.

Another common type of carbide is the $M_{23}C_6$ type. The primary metallic constituent is chromium with small amounts of molybdenum or, when present, iron. A typical composition is $Cr_{21}Mo_2C_6$. Generally formed in the grain boundary areas, this carbide may assume one of two morphologies, Figure 5. If discrete particles of the carbide are formed, creep resistance and ductility are enhanced, especially in moderate temperature ranges. However, in certain instances a continuous carbide film or platelet is formed in the grain boundary regions. This network has an embrittling nature and a deleterious effect upon both creep resistance and ductility. The morphology of grain boundary $M_{23}C_6$ is controlled through composition, processing, and heat treatment.

Generally, the least common of the three major carbides detected in nickel-base superalloys is the M_6C type. A more correct designation is MxMyC where x + y = 6. Mx is generally one of the transition metals (Fe,Ni, or Co) and My is one of the refractory metals (Mo,W, or Cb). The composition of this carbide can vary widely with the composition of the alloy. Alloy chemistry also greatly affects the amount of this carbide formed. It is usually more stable than $M_{23}C_6$, but somewhat less so than MC. M_6C assumes a rather small discrete particle morphology, often in grain boundary regions.

2. Wrought Alloys

From the development of the early Nimonic alloys on through the introduction of Udimet 700 in 1959, nickel-base superalloys were either exclusively wrought or were combination wrought-cast alloys. The uniformity of properties and structure, reliability, and lower cost of production of high temperature products forged from wrought nickel-base alloys was unquestioned. In fact, these alloys were used in the as-cast form only when the part could not be produced by conventional forging techniques.

The chemical composition of a few common wrought nickel-base alloys are shown, Table I, and are designated by "W". The increasing amount of solid solution and intermetallic forming alloying elements present as one progresses from Nimonic 75 to Udimet 700 is apparent. The improvement in strength with increased alloy additions is shown in Table II, which gives the 1800°F and 1900°F stress rupture properties. Alloys such as Nimonic 75, Inconel X, and Waspaloy are not used at temperatures as high at 1800°F. To achieve the necessary strength levels, wrought superalloys require complex heat treatments consiting of solution annealing and often several aging cycles.

In addition to the high temperature strength requirement, wrought alloys must also possess the ability to be deformed at some temperature above the potential use temperature. Demands for increased strength were fulfilled by increasing the amounts of alloying constituents. While additional alloying provided the required strength, it also rendered nickel-base alloys stronger at the potential deformation temperatures. As a result, wrought alloys became more difficult to produce as bar stock and forge to finished parts.

3. Cast Alloys

The advantageous position enjoyed by wrought superalloys in relation to cast alloys has today been all but nullified when strength retention above 1800°F is demanded. Casting practice has improved tremendously to the point where uniformly sound, reliable products are often the rule. The advantages of cast alloys as creep resistant materials above 1800°F are two-fold: (1) a given alloy, in the cast form is more creep resistant than in the wrought form, probably because of grain size and subgrain effects; (2) increased amounts of aluminum and titanium and the solid solution strengthening refractory elements can be added with the removal of the workability criterion.

Typical compositions for a number of currently prominent cast nickel-base superalloys are shown in Table I. While increased amounts of alloying elements are evident when compared with the wrought alloys, the trend is also apparent from an earlier cast Inco 713C to the recently developed NASA alloy TAZ8. The improved strength derived from these alloy additions is presented in Table II.

A further advantage of cast alloy materials is the ability to be precision cast to hollow configurations. Providing that mold technology and alloy castability is sufficient, a wide variety of intricate parts can be investment cast by the lost wax method.

B. New Alloy Requirements

1. Mechanical Properties

In order to satisfy the requirements for a turbine blade material in new higher temperature gas turbine engines are described previously, a substantial increase in the presently available properties of nickel-base superalloys is necessary. Certain guidelines must necessarily be established as target properties for the new material. Specifically, the target of this program is the attainment in a nickel-base superalloy of 3000 hour stress rupture life in an 1875°F test under 15,000 psi load. This requirement represents a considerable increase over properties available in present-day cast superalloys.

Properties other than stress rupture life are also of varying importance in a nickel-base superalloy and reflect the ability of the alloy to perform under a variety of potential conditions. Satisfactory tensile strength is required at all temperatures below the projected use temperature. Adequate tensile and stress rupture ductility are important to safeguard against brittle behavior. The ductility requirement is quite often a thorny problem, especially in the range of 1400°F where a minimum is observed to occur in many creep resistant nickel-base alloys.

2. Non-Mechanical Properties

Certain additional features which a nickel-base alloy must possess should also receive attention. They are not mechanical properties and therefore do not receive primary consideration but are nonetheless important and may serve to differentiate between two alloys which exhibit equal mechanical properties.

a. Castability

Adequate castability is a prime requisite in any cast nickel-base superalloy, especially if complex cooling passages are to be incorporated into the design. Specifically, the following should receive at least qualitative attention: resistance to hot tearing and hot shortness; minimum dimensional changes upon cooling; controllability of grain size; and resistance to the formation of microstructural inconsistancies, such as primary gamma prime.

b. Workability

Should it be desired to use the alloy in the wrought form, a degree of workability is necessary. In most nickel-base superalloys a rather narrow working range exists, bounded on the high side by a "hot shortness" condition at grain boundary areas and on the low side by excessive strain hardening and cold cracking. Generally, the proper range for the most highly alloyed materials is in the range from 2000°F to 2100°F. Close control of reductions, temperatures, and microstructure usually aid in the solution of the workability problem.

c. Thermal Fatigue

Thermal fatigue cracking is a major cause for rejection of turbine blade materials. It results from the alternating tensile-compression stress condition produced in a constrained part subjected to thermal expansion and contraction. Such factors as thermal conductivity and expansion, geometry, nature of the thermal cycle, and environment all affect the thermal fatigue behavior of a metal[7].

d. Oxidation

The resistance to oxidation of most nickel-base superalloys is attributed to the formation of a spinel which is relatively continuous, adherent, and impervious to either oxygen or metal diffusion. The addition of chromium and aluminum is essential to the oxidation resistance of these materials. At 1875°F, in an oxidizing atmosphere, oxidation of nickel-base superalloys is probably not of major concern, particularly if a protective coating is used.

e. Structural Stability

A characteristic of nickel-base superalloys, common to many alloys, is the metastable nature of the phases present. At elevated temperature, the rates of change for various phases are increased and the stability of the strengthened alloys becomes a problem. The morphology of the gamma prime precipitate is vastly different at high and low temperatures. The carbide phase $M_{23}C_6$ is believed to decompose in most alloys at temperatures below 1875°F. (Typical is the case of Rene 41 in which $M_{23}C_6$ has been determined to decompose somewhere below 1900°F[8]).

C. Theoretical Basis

Previously discussed were the general reasons for the selection of the approach to be taken in developing a material capable of extended service in new, hotter gas turbine engines. Obviously, the upgrading of present-day nickel-base superalloys will require the use of all three available strengthening mechanisms: intermetallic precipitation; solid solution strengthening; and carbide formation. Since the alloy to be developed is intended for use primarily as a cast material, it is important that the phases which contribute to these phenomena be distributed in a desirable manner directly upon solidification. The basis for the usage of each mechanism will next be described, outlining the ways in which it will be optimized and what role it contributes to the overall strength of the alloy.

1. Intermetallic Precipitation

Nickel-base superalloys represent a classic example of the strengthening imparted to a metal by the precipitation of a second phase; in this case the intermetallic Ni_3Al, commonly referred to as gamma prime. A certain degree of alloying can occur in this compound as a considerable portion of the aluminum can be replaced with titanium. Columbium and cobalt may also alloy with gamma prime.

When used in the cast form, the gamma prime configuration which produces strengthening is actually a partially overaged structure, having a cubic form (Figure 2A). The lattice constants of gamma prime are quite similar to those of the nickel-rich solid solution or gamma phase. Because of this observation, it might be assumed that the major strengthening effect of the precipitate, while probably coherent with the matrix is not due to lattice straining, but to a variation in flow characteristics between the two phases. In fact, it has been demonstrated that the flow stress of the gamma prime phase increases with temperature up to 1400°F[9]. Thus, the precipitated phase actually becomes more resistant to deformation while the matrix loses strength. (Fine second generation gamma prime is often encountered in heat treated superalloys following aging treatments in the 1500°F to 1700°F range. This precipitate apparently enhances lower temperature strength but is of little value as a strengthener above 1600°F).

Undoubtedly, gamma prime is a very effective strengthener at intermediate temperatures. However, under the required operating conditions, intermetallic strengthening may be of marginal value, especially over extended time periods. Gamma prime will provide a considerable measure of strength to the alloy in the intermediate temperatures, and its benefit must be used fully in any new composition. Furthermore, it remains probable that there is a significant strengthening contribution from Ni_3Al up to 1900°F or more, particularly if a proper phase morphology can be established.

The most recent cast nickel-base superalloys have contained total amounts of aluminum and titanium in excess of 5 percent by weight. Most often, the proportion of aluminum exceeds that of titanium (see Table I). This has the added advantage of increasing the oxidation resistance of the alloy.

The proposed plan of attack is to treat the gamma prime phase as a prime strengthening mechanism, investigating fully the most promising composition ranges for the intermetallic formers, aluminum and titanium.

2. Solid Solution Strengthening

It has been shown above that the precipitation of intermetallic gamma prime is responsible for a diminishing portion of the total superalloy strength as operating temperatures exceed 1700°F. Matrix strength, thus, becomes of greater importance. All nickel-base superalloys receive some solid solution strengthening effect from the assortment of addition elements in the composition. Reference to the table of compositions (Table I) reveals that although the proportion of the elements added specifically for solid solution strengthening (generally the refractory metals) increases in the newer higher strength alloys, usually fairly large amounts of only one or two such elements are present. While this procedure has obviously resulted in a large increase of stress rupture life, still greater improvement is needed in operational temperature into the range where solid solution strengthening may very well be of primary importance. As a means to accomplish this improvement, the hypothesis of Guard will be used[10]. This theory holds that the first incremental addition of an alloying element has a greater effect than each successive increment.

In order to accomplish this objective of increased temperature capability through solid solution hardening, the proposed addition of varying amounts of the three highest melting refractory metals tungsten, tantalum, and molybdenum will be investigated. The levels to be examined are in the range from 1 to 10 percent. There are other elements potentially useful for solid solution strengthening in nickel-base superalloys. Among them, the precious metals of the platinum group and rhenium have recently received attention[11,12]. Cost, availability, and potential merit as a strengthener indicates that ruthenium should be the most useful platinum group element. Therefore, ruthenium and rhenium solid solution strengthening will be evaluated.

3. Carbide Formers

The importance of carbide structures in the proper relationships to the strengthening of nickel-base superalloys has been previously indicated. The carbon content of most nickel-base alloys is around 0.10 percent. TRW internally conducted investigations have shown a large deterioration in high temperature properties, particularly ductility, and workability when carbon levels have been reduced in several different materials. Reference to Table I reveals that chromium and titanium are the carbide forming elements commonly added. Columbium, tungsten, and molybdenum, while present in several of the alloys for solid solution strengthening are moderately potent carbide formers. Other elements occasionally added which have varying abilities to combine with carbon in superalloys are tantalum and vanadium.

In order to derive the maximum benefit from proper carbide formation, effort will be made to optimize the proportions of several strong carbide forming elements. While chromium, titanium, tantalum, tungsten, and molybdenum will be added for other reasons, they also should contribute to the formation of a desirable carbide dispersion. There are, however, other elements which can be added specifically to produce thermally stable carbide phases. In the present investigation, hafnium, vanadium, and columbium three carbide forming elements, will be added specifically to improve the high termperature stability of various types of carbides. It is expected that rather small amounts of each will prove most desirable (not more than 2 percent). The concentrations of these three elements will be optimized. It is expected that small additions of two or three of these elements will prove more advantageous than a large amount of one.

4. Minor Elements

Minor elements purposefully added to nickel-base superalloys are boron, zirconium, and carbon (the role of which has previously been treated[13]. Zirconium and boron were orginally present as tramp elements. However, attempts to lower their concentration or eliminate them completely has in most cases led to a deterioration in properties. Present-day alloys, therefore, usually contain small amounts of both. Boron content ranges from .005 percent to .03 percent; with approximately the same amount of zirconium. The exact function of these elements is not completely established; however, it is felt that they retard segregation of other important ingredients in grain boundary areas. The exact amount of born and zirconium required for maximum high temperature properties will be determined in the final alloy composition after the other constituents have been fixed. However, a percentage of each, consistent with current practice, will be maintained in all experimental compositions while the final composition is being determined.

D. Evaluation Procedures

A variety of testing procedures are required to fully evaluate a material for elevated temperature blading applications. These include not only the normal tensile and stress rupture tests, but also thermal fatigue and oxidation tests. However, for screening purposes, only stress rupture and tensile tests are programmed. These tests and the statistical procedure used to analyze them will next be described. Workability tests will also be performed on each alloy. Castability will be observed.

1. Stress Rupture Testing

As has been previously suggested, the stress rupture test is the most practical method of evaluating potential high temperature materials. In the present investigation, objective properties include 3000 hour life at 1875°F and 15,000 psi load. Because of the excessively long times indicated by these test conditions, expediency dictates a shorter test for the alloy survey portion of the project. For this purpose, the temperature parameter has been changed so that much shorter test times will be evolved; the actual test will be conducted at 2000°F under the same 15,000 psi loading. When tested under these conditions, lives from 50 to 150 hours can be considered to roughly correlate with the 3000 hour criterion at 1875°F in a highly alloyed nickel-base superalloy similar to MAR M-200 or IN-100, Figure 6[14].

It should be borne in mind that certain metallurgical changes quite possibly occur in the temperature range from 1875°F to 2000°F. Particularly, the gamma prime strengthening mechanism which is vital to intermediate temperature strength in nickel-base alloys may lose its potency at 2000°F. Thus, an overemphasis might be placed on high temperature strength, probably derived primarily from solid solution hardening, by the very nature of the test. It is felt, however, that even at 1875°F the value of gamma prime as a strengthening agent will be somewhat diminished, especially over long periods of time, and that the 2000°F test may well be a valid indicator.

2. Tensile Tests

Some further insight into the performance of a potential alloy candidate at both 1875°F and 2000°F can be obtained from the short time tensile test. It is also important to determine the tensile behavior of the material at lower temperatures, ambient temperature and 1400°F are typical evaluation points. These tests reveal the maximum strength inherent in the alloy and the ductility it possesses. Ductility is particularly important in the 1400°F temperature range where many nickel-base superalloys undergo embrittlement in service.

3. Analysis of Data

In order that the maximum amount of information be obtained from as few experimental heats as possible, an adequate statistical design is desired to analyze the data. For this purpose, the Latin Square design will be utilized[15]. This technique is particularly useful where variables can be separated into groups of three, ideally having a similar function. This is often the case in nickel-base superalloys. Each of the three elements in a given grouping are assigned three compositional levels for a total of nine alloys for each square. By analyzing the sum of squares and using the "F" test for significance, the individual consequence of each element can be ascertained and the interaction between the elements can be scrutinized. One is then able to discern the level of confidence in certain data trends. For example, when varying the element molybdenum over a given range using three values, it can be determined whether or not this element has a significant effect upon the property changes which may develop as the chemistry is altered.

4. Workability Testing

In gaging the potential usefulness of a new nickel-base superalloy as a wrought material, some test for workability must be conducted. At TRW, both side pressing and rolling tests have been successfully used to qualitatively measure this property. Such a test not only rates the relative workability of various superalloys, but also aids in the determination of a working temperature.

5. Structural Evaluation

A variety of photographic techniques aid the physical metallurgist in the explanation of observed phenomena. The following will be employed as necessary:

a) Photomacrographs to allow a study of the fractured area in broken specimens. Brittle and ductile behavior can be studied.

b) Microscopy to evaluate the as cast structure microstructure of the alloys. The various phase morphologies will be studied with both light and electron metallographic techniques.

c) Phase analysis on selected alloys to determine the carbides present. This technique involves the X-ray examination of electrolytically extracted residues.

In later portions of the work attempts to alter the structure of wrought alloys will be made through the use of appropriate heat treatments. These same metallographic and extraction techniques will be used in the case of both cast and wrought alloys.

III MATERIALS AND PROCEDURES

The casting and evaluation techniques utilized in the development of nickel-base superalloys in this project did not involve any radical departures from normal techniques employed for this type of material. Vacuum melted tensile specimens were tested and evaluated using statistical techniques. The resulting test data were used to formulate a base composition for further alloying studies. These test procedures will be discussed in the following sections.

A. Materials

In order to minimize undesirable side effects from contaminating elements, raw materials of the highest purity which could be economically obtained were utilized. A purity level of 99.9 percent was specified on most ingredients. Master alloying constituents were also acceptable if the total impurity level as determined by the desired alloy composition did not exceed 0.1 percent. Those alloying materials which were not readily obtainable at the required purity level were used at the following levels:

Constituent	Minimum Purity
Aluminum	99.8%*
Chromium	99.5%
Cobalt	99.5%
Vanadium	99.7%
Titanium	99.8%
Boron (NiB)	99.0%

* For the first 27 heats the minimum purity of aluminum was 99.7%.

B. Casting Procedure

Alloy preparation was conducted in either an NRC or a Stokes vacuum induction melting furnace. These units are fitted with 50 pound crucibles and are capable of maintaining vacuums of less than 10 microns of mercury. All melting and casting operations were closely monitored by qualified technical personnel fully acquainted with the objectives of the program. The procedure used for each alloy was basically a two step operation. This consisted of meling virgin alloying ingredients and casting the metal into a cylindrical shell mold. The ingot so produced was remelted and poured into pre-heated tensile-bar cluster molds. The Series I melting procedure utilized for the first step is outlined in Table III. Modifications of the basic sequence necessitated by the additional ingredients in the Series II and Series III compositions are also noted.

A basic alloy composition was selected for all Series I compositions. It is given below in weight percentages:

Co	Cr	Zr	B	C	Ni
10.0	10.0	0.03	0.02	0.13	Bal.

Within the limits of this base the following elements were varied between the limited shown:

Mo	W	Ta	Al	Ti
1.0-8.0	1.0-10.0	1.0-8.0	4.5-6.3	1.0-1.8

Following the analysis of Series I data the following fixed additions were made to the base shown above:

Mo	W	Ta	Ti
2.0	5.5	8.0	1.0

Using the new base and 4.5 percent aluminum the following elements were varied in Series II:

Cb	V	Hf
0-2	0-2	0-2

Series III alloys evaluated variations in the elements listed below:

Al	Re	Ru
4.5-6.3	0-4	0-2

The total weight of charged materials was 20 pounds for series I alloys. An increase in the charge to 24 pounds was incorporated for Series II and III alloys. Of this weight approximately 2.5 to 3 pounds was required for chemical analyses, the remainder being used in the cluster mold. Table IV presents the procedure followed when the ingots were remelted and cast into tensile bars. Pertinent melting data for both portions of the procedure are summarized in Table V. The tensile bar cluster which was produced is similar to that illustrated in Figure 7. This configuration yields fourteen tapered .250 inches gage diameter tensile bar preforms (minimum diameter approximately 0.275"), two flat sections, two square and two round bars, and a 1 inch diameter "down pole" about 3 inches in length.

Temperature measurements were conducted with Pt-Pt13%Rh thermocouple which could be submerged into the molten bath when desired. A vacuum seal enabled thermocouple changes to be made while a heat was in progress. The approximate temperature at which the onset of solidification occurred was estimated by observing the point at which small metallic crystals (termed "silver-fish") began to collect in the liquid bath.

Since certain elements are to be evaluated at the zero concentration level in Series II and III alloys, extreme care was necessary so that contamination did not occur. This was accomplished by carefully planning the sequence of melting and by performing "wash" heats at required points in the sequence. The "wash" heats were of pure nickel and served to scavenge all potential contaminants.

Variations in the time required for a heat were primarily due to differences in the melting time and pump down time. When the furnace was used after a period of inactivity, achievement of the proper vacuum was always slower. Also, when the crucible had been used previously and was thoroughly heated, melt-down progressed more rapidly. Further variations were attributed to variations in the duration of the carbon boil. This phase of the melting sequence was carried to completion in every case, but times ranged from five minutes to about fifteen. The effectiveness of the boil and the highly purified starting materials were attested by the extreme cleanliness of all heats.

On the remelt portion of the casting procedure the times were understandably much shorter than those for the virgin melting. In the time used (less than 30 minutes) the loss of heat by the preheated cluster mold was at a minimum. While accurate temperature measurements were not available, the mold still emitted a full red-orange glow. It was therefore estimated that the temperature was near 1500°F. This mold temperature effectively retards solidification rate to a point where sound, shrink-free castings were obtained with little segregation. Remelting was conducted in a ceramic crucible liner which was not reused.

C. Inspection

A complete chemical analysis for all alloying elements and important impurities was conducted for each heat. The results are tabulated for the 27 Series I alloys, Table VI, together with the "aim" analysis. (The analysis for Series II and III heats are incomplete at this time, although the aim chemistries are shown in Table VII.) Because of the absence of X-ray or spectroscopic standards the determinations for most elements were performed by wet methods. These results indicate the generally excellent recovery of alloying elements obtained by the melting and casting procedure utilized. Only in the aluminum content of Alloy XII is there a serious discrepency between the aim analysis and the actual result. It is suspected that this error can be attributed to the loss of a portion of the aluminum addition while charging it to the molten bath. The extremely low levels detected for the impurity elements; silicon, iron, sulfur, and maganese in the alloys thus far tested, reflect the high quality of the starting materials and the well established melting practice.

Non-descructive testing was conducted on all test bars prior to and after machining. The soundness of all cast preform tensile test bars was examined by radiography. In all cases a sufficient number of completely sound bars were available for subsequent testing. After machining, all finish-ground specimens were tested by the fluorescent dye pentrant technique. Of 324 specimens prepared only two were rejected because of imperfections.

D. Mechanical Testing

Two types of mechanical property tests were performed on the cast nickel-base superalloys. Stress rupture testing was conducted in air according to ASTM standards on Satec constant load, lever arm machines. Temperature control was maintained to within ±3°F. High temperature tensile testing was conducted on a constant crosshead speed Instron Universal Testing Machine at a speed of 0.020 inches per minute. A Baldwin Testing Machine was used for the majority of the room temperature and 1400°F tests, the remainder being conducted on the Instron unit. A strain rate of 0.005 inches per inch per minute as measured with a strain pacer was utilized for testing on the Baldwin machine. There was no obvious variation between tests run on the Baldwin and those conducted on the Instron. ASTM standards and furnace control to within ±3°F were adhered to. Duplicate specimens were run on all tensile tests. Some indication of the gage length is required on all specimens tested on the Baldwin apparatus. For this purpose, one specimen was marked with small gage points separated by 1 inch. The radius-to-raius distance was used as a gage length in the second specimen. There did not appear to be any significant difference between the two methods and, therefore, little effect caused by the superficial gage marks.

The data obtained from these tests have been analyzed both by simply averaging results for different compositional levels and by the Latin Square method. A sample calculation is shown in the Appendix. Stress rupture life data times 10^2 were converted to common logarithms prior to analysis while all other data were treated without alteration. The factor of 10^2 was used for ease of calculations. All three samples of Alloy I broke upon loading at 15,000 psi and 2000°F. They were arbitrarily assigned the relatively small value of 0.01 hours for stress rupture life. All times over 30 minutes were recorded on timers to tenths of hours. Times shorter than 30 minutes were recorded with a stop watch and converted to hundredths of an hour.

E. Metallography

In order to observe the microstructure of the various cast nickel-base superalloys, a transverse sample was cut from the threaded end of a room temperature tensile bar and prepared for metallographic observation by mechanical polishing. The chemical etching reagent used consisted of 62% H_2O, 15% HF, 15% H_2SO_4, and 8% HNO_3. Photomicrographs of each of the 27 Series I alloys were taken in the etched condition at both 250X and 750X magnifications. Details of the fractured area in stress rupture tests were also examined. Macrophotographs were made at 5X magnification of the fracture area from a representative specimen of each alloy. This type of a photograph permits the observation of secondary cracking and the relative ductility or brittleness of the fracture. Microscopic fracture characteristics, such as whether it is of intergranular or transgranular nature, were studied at 60X magnification on representative stress rupture specimens. All alloys were photographed.

F. Workability Testing

The workability of the various cast alloys was gaged by extrusion of the "Down pole" portion of the tensile bar cluster and side pressing to 50 percent reduction of the resulting stock. The down poles were X-rayed for porosity prior to extrusion. Billets were then prepared for extrusion by "conditioning" their surfaces, finish machining to approximately 0.900 inch diameter, and cladding in mild steel to 1.500 inch diameter. The billets were then extruded at 2075°F to a 4:1 reduction in cross-sectional area. Extrusion was accomplished on a vertical hydraulic press of 150 ton capacity. It is anticipated that this procedure will yield at least 6 inches of sound 0.450 inch diameter bar stock. The extruded material will then be sectioned into 1 inch long pieces and side pressed to a 50 percent diameter reduction at various temperatures in the range 2000°F to 2200°F. Side pressing will be conducted between flat dies on a hydraulic press.

IV RESULTS AND DISCUSSION

A. Series I Alloys (27 Compositions)

The evaluation of castability and mechanical properties of the first group of alloys has been completed. Light microscopy is also complete. Workability testing has been initiated.

1. Castability

The castability of the series of nickel-base superalloys developed in Series I can be rated by several qualitative means. Reports of the technical monitors present during all melting operations indicated that the alloying ingredients were easily melted within the limits of the established procedure. Reactive metals were added without violent reactions and metal loss. Fluidity at the pouring temperature was excellent. All heats were remarkably free of impurity slags. The recovery of alloying elements is generally quite good as was shown in Table VI. It is felt that the only notable variation between the aim and the actual analysis was in Alloy XII where the aluminum content was low.

Further verification of the excellent castability of these alloys is the complete absence of porosity detectable by standard radiographic techniques in the tensile specimens. There were no rejections among the 27 heats for this reason. Certain of the "down poles" subsequently extruded for workability studies did contain a small degree of porosity near the upper end as revealed by X-ray inspection. However, these portions were largely removed prior to extrusion. Rejection of machined specimens in the dye pentrant test was also minimal. Only two finish ground test bars were withheld from testing for small imperfections. There was no evidence of cracking or surface checking from machining and grinding operations. It can, therefore, be concluded that any one of the Series I alloys fulfills the requirement of adequate castability and also possesses sufficient machinability.

2. Mechanical Property Tests

Mechanical property test results were used to determine elemental variation effects in the base composition:

Co	Cr	Zr	B	C	Ni
10.0	10.0	0.03	0.02	0.13	Bal.

It was on the basis of these tests, particularly 2000°F stress rupture tests, that a selection was made of the base composition for Series II and III studies. While all 27 alloys were considered individually, they were grouped by common Al and Ti levels for the Latin Square portion of the analysis. The grouping utilized is shown:

Latin Square 1

4.5Al - 1.0Ti

Vary Mo - 1, 4.5, and 8

Vary W - 1, 5.5, and 10

Vary Ta - 1, 4.5, and 8

Latin Square 2

6.3Al - 1.0Ti

Same Mo, W, and Ta variations as for Latin Square 1.

Latin Square 3

6.3Al - 1.8Ti

Same Mo, W, and Ta variations as for Latin Square 1.

Thus, it was possible to obtain a statistical analysis of the effects of Mo, W, and Ta in each of three base compositions containing different Al and Ti levels.

a. Stress Rupture Properties

The results of 2000°F stress rupture tests conducted under 15,000 psi loading are shown in Table VIII. The generally excellent reproducibility obtained for the triplicate tests on a given alloy indicates again the integrity of the castings. A statistical evaluation of the 2000°F stress rupture life for each aluminum and titanium level are shown in Table IX. Experience has indicated that time parameters, such as stress rupture life, should be converted to logarithmic functions to obtain more reliable data correlations than are obtained with raw data. This is in agreement with the Larson-Miller parameter which uses a logarithmic function of time to correlate time, temperature, and stress level in stress rupture and creep tests.

The average stress rupture life (actual) for each element varied has been plotted as a function of the percentage of that element contained for each base Al-Ti level, Figure 8. All elements were found to exert a significant influence upon stress rupture life at 2000°F in a base containing 4.5Al-1.0Ti. However, only tantalum significantly affected the properties in a 6.3Al-1.0Ti alloy, and none of the refractory elements investigated were statistically significant in 6.3Al-1.8Ti composition. All significant data show that an increase in the amounts of Mo, W, or Ta is beneficial to the 2000°F stress rupture life of the alloy tested. Of the three, tantalum appears the most effective, with a maximum average life of nearly 10 hours for 8%Ta alloys containing 4.5Al-1.0Ti. Tungsten appears to be of more value as a strengthener than does molybdenum. It seems apparent from these data that the 4.5Al-1.0Ti alloy is the most effective base when the percentages of refractory additions are high. The 6.3Al-1.0Ti base is the next most effective, when tantalum is high and tungsten and molybdenum are at intermediate levels. The third base, 6.3Al-1.8Ti, exhibits the poorest properties of the three.

Stress rupture ductility was measured by the percent elongation and the percent reduction in area criteria. For this analysis, percent elongation was evaluated by statistical methods. Data, obtained from Table VIII, have been analyzed by the Latin Square Method, Table X. Results are shown graphically in Figure 9. As might be anticipated, the addition of the refractory metal solid solution strengtheners, Mo, W, and Ta, generally serves to decrease the stress rupture ductility. In the 4.5Al-1.0Ti base alloy, all three additions have a significant effect. Both Mo and Ta are significant in the 6.3Al-1.0Ti alloy, while only W is of significance in the 6.3Al-1.8Ti base.

While it is apparently desirable to incorporate large amounts of at least Ta and W from a standpoint of stress rupture life, ductility measurements indicate the opposite. Thus, a potential idealized composition for use as a future base might contain 8.0%Ta, 5.5 to 10.0%W, and 1.0 to 3.0%Mo. Since it is desirable from a density standpoint to minimize the amount of tungsten in the alloy, an 8.0%Ta, 5.5%W, 2.0%Mo alloy appears to be the most desirable composition for future study.

The selection of a proper Al and Ti level for future work is less obvious. Titanium is most profitably set at 1.0 percent, but the data can be used to support the selection of either 4.5%Al or 6.3%Al alloys. As a result of this uncertainity, it was concluded that aluminum should be varied between 4.5 percent and 6.3 percent as a portion of a future Latin Square analysis.

While the primary purpose of a Latin Square is to statistically analyze the effects of given elements, certain auxiliary benefits can be derived from test results. Examination of the test data, Table VIII, reveals that alloys V, VII, and XIII have particularly good stress rupture lives.

Alloy	Life	Mo	W	Ta	Al	Ti
V	11.9 hours	1.0	5.5	4.5	6.3	1.0
VII	16.4 hours	1.0	10.0	8.0	4.5	1.0
XIII	8.5 hours	4.5	5.5	8.0	4.5	1.0

Since all other compositions had lower values when tested, the merits of an 8%Ta, 5.5%W, 2%Mo base alloy can be seen. These data also substantiate the selection of 1.0%Ti and the decision to continue the variation of aluminum further. Some indication of the strength of these alloys in relation to contemporary cast nickel-base superalloys can be gained by comparison of stress rupture properties. Figure 6 indicates that MAR-M-200, for example, can be typically expected to provide about 15 hours of life under the test conditions used in this investigation. Thus, Alloy VII, in particular, compares favorably with the best of the present-day commercial cast alloys.

b. Tensile Properties

Tensile test results for 2000°F and 1875°F tests are shown in Tables XI and XII respectively. Latin Square data analysis is summarized in Tables XIII and XIV for ultimate tensile strength and Tables XV and XVI for percent elongation. Figures 10 and 11 graphically show the tensile strength data while Figures 12 and 13 show ductility results. These results generally confirm what would be expected, that there is a significant loss in strength between 1875°F and 2000°F. However, this trend appears uniform in all alloys and it can, therefore be surmised that no changes in strengthening mechanism are occurring in the leaner alloys which do not take place in the more heavily alloyed compositions. It is also of note that the three alloys which were selected previously as possessing superior stress rupture properties V, VII, and XIII also possess among the best tensile properties at both 1875°F and 2000°F (about 50,000 psi and 20,000 psi respectively) with better than average ductility. Thus, the 2000°F, 15,000 psi survey test appears to be a reasonably valid substitution for the final target criterion at 1875°F and 15,000 psi.

Considering the Latin Square analytical data, (Tables XIII to XVI and Figures 10 to 13) it is found that all three refractory metal additions Mo, W, and Ta have a significant effect upon tensile strength at both 2000°F and 1875°F in a 4.5Al-1.0Ti base. As the base composition is altered to include larger percentages of Al and Ti, it is found that only Ta has a significant effect and it is effective only in the 6.3Al-1.0Ti base tested at 2000°F. As might be surmised from a knowledge of the stress rupture test results, tantalum has the greatest effect upon 2000°F and 1875°F tensile strength, being most effective at the 8.0 percent level. Tantalum is most effective in the 4.5Al-1.0Ti base and nearly as potent in the 6.3Al-1.0Ti alloy. Tungsten and molybdenum follow in decreasing order of effectiveness.

The effect of refractory metal additions upon tensile ductility at 1875°F and 2000°F was gaged by statistically analyzing percent elongation test data. It was found that, as with the tensile strength, all three addition elements were significant when employed in the 4.5Al-1.0Ti base at both test temperatures. Also tantalum was significant in the 6.3Al-1.0Ti base and tungsten in the 6.3Al-1.8Ti alloy when tested at 1875°F. The general effect of increasing the amount of any refractory metal addition was to reduce the ductility of the alloy.

Just as the 1875°F and 2000°F tensile tests can be considered together, the nickel-base alloys tested at room temperatures and at 1400°F exhibit similar behavior. These properties are tabulated in Tables XVII and XVIII for 1400°F and room temperature tests respectively. In these temperature ranges, especially at 1400°F, the property of most concern in nickel-base superalloys is generally the ductility. Scrutiny of the data shows that ductility is indeed a problem. Certain alloys exhibited no 0.2 percent offset yield point and less than 1 percent elongation and area reduction. However, it is encouraging to note that the alloys observed to possess the best stress rupture properties, alloys V and VII, also display better than average ductility at 1400°F and room temperature.

The Latin Square statistical data for 1400°F and room temperature tensile strength are shown in Tables XIX and XX. Results of the analysis are illustrated graphically in Figures 14 and 15. The same trends explained previously for higher temperature tests seem valid in these cases also. All refractory metal additions have a statistically significant effect upon tensile strength in the 4.5Al-1.0Ti base alloy (except Mo at room temperature). The effects in bases containing higher Al and Ti contents are generally not significant. It is of interest to note that in some cases the strength of an alloy is greater at 1400°F than it is at room temperature (Alloy V for example), perhaps because of the increasing flow stress of gamma prime with temperatures up to about 1400°F[9].

Tables XXI and XXII give pertinent statistical data used in the calculation for significance of the 1400°F and room temperature elongation values. Figures 16 and 17 are graphical representations of this data. In these cases, an opposite effect appears from that which was manifested in previous Latin Square calculations. Refractory metal additions were deemed significant in the two bases containing higher Al and Ti levels. Molybdenum and tungsten also significantly affected the room temperature elongation in the 4.5Al-1.0Ti base alloy. Increased amounts of both molybdenum and tungsten resulted in a decrease in elongation at both test temperatures. However, ductility was at a maximum when the tantalum content was at the middle level.

3. Metallographic Analysis

The mechanical property results previously discussed have served to differentiate between the various superalloy compositions which have been evaluated. Within the compositions tested, there is a wide variation in the properties encountered. Some alloys exhibit a high degree of ductility and little strength, while others possessed moderate strength but were extremely brittle. The most promising compositions displayed not only higher strength, but also a moderate degree of ductility. The effects of increasing the amount of refractory elements and altering the proportions of aluminum and titanium upon the mechanical properties of an experimental series of nickel-base superalloys can, thus, be statistically studied. It has been found that significant amounts were detrimental, especially when the proportions of aluminum and titanium were high. Metallographic studies can be used to help determine the phases present in these alloys so that a correlation can be made between microstructure, properties, and chemistry.

a. Cast Condition

The initial condition for this investigation, and that in which a new nickel-base superalloy will most likely be used, is the as-cast condition. Therefore, initial microstructural evaluations were conducted on this structure. The total amount of solid solution elements varies widely within the 27 Series I compositions. The percentage for each group of three (having increasing amounts of aluminum and titanium within a group) are shown below:

I, II, and III 3%	X, XI, and XII 10%	XIX, XX, and XXI 17%
IV, V, and VI 11%	XIII, XIV, and XV 18%	XXII, XXIII, and XXIV 14.5%
VII, VIII, and IX 19%	XVI, XVII, and XVIII 15.5%	XXV, XXVI, and XXVII 22.5%

There is obviously a wide variation of the total amount of refractory metal additions, ranging from 3 percent to 22.5 percent. It would seem likely, therefore, judging from the mechanical properties, that such a large disparity would have an apparent effect upon the microstructural balance of the alloy. Also having a great influence upon mechanical properties, and thus presumably upon microstructure, is the variation of gamma prime formers, aluminum and titanium.

Some of the various gamma prime formations encountered are shown in Figure 18. The microstructure of Alloy II containing only 3 percent total refractory metal additions and 6.3%Al-1.0Ti is illustrated in parts A and B. This microstructure is also typical of other especially lean alloys. A carbide network is observed to outline grains. When viewed at a higher magnification, Figure 18B, the lamellar nature of the carbide phase can be observed. This particular sample has been lightly etched. However, it is still possible to discern the gamma prime formations within each grain. This phase has probably been formed upon cooling of the solid solution in the range of 2100°F to 1900°F. Figure 18C illustrates the slightly more populous gamma prime formations encountered in Alloy III, which is identical to Alloy II except that the titanium level is now 1.8 percent. Since this sample has been more heavily etched than the previous one, an effect which tends to emphasize the gamma prime particles, the variation in texture of this phase throughout a grain is evident. In certain areas, the suggestion of lamellar formations of gamma and gamma prime are present. This sequence, carried one step farther, is shown for Alloy VI in Figure 18D. In this micrograph, massive gamma prime formations are prominent. These structures may be of the type previously shown at 3400X magnification, Figure 3. Alloy VI, like Alloy III, also contains a 6.3%Al and 1.8%Ti. However, the total refractory metal percentage has been changed from 3 percent to 11 percent by increasing Ta from 1 percent to 4.5 percent and W from 1 percent to 5.5 percent. Apparently, larger amounts of refractory metal alloying elements, when present with sufficient aluminum and titanium (i.e., at least 6.3%Al and 1.0%Ti) encourage the formation of massive gamma prime, an effect analogous to lowering the ratio of Ni to Al and Ti. However, the additional benefit derived from the refractory metal solid solution strengthening is greater than any detrimental effect caused by eutectic gamma prime. This can be demonstrated by comparing the average stress rupture life of Alloy III (0.11 hours) with that of Alloy VI (7.8 hours).

When discussing the relative merit of the various alloys tested, it was pointed out that Alloys V, VII, and XIII possessed properties superior to the others. Pertinent compositional variables are as follows:

Alloy	Total Refractory Metal	Mo	W	Ta	Al	Ti	Average 2000°F 15,000 psi Stress Rupture Life
V	11%	1%	5.5%	4.5%	6.3%	1.0%	11.9 hours
VII	19%	1	10.0	8.0	4.5	1.0	16.4
XIII	18%	4.5	5.5	8.0	4.5	1.0	8.5

While the strengths of Alloys V and VII are similar, it is apparent that it was derived in a different manner in each. Alloy V has a higher percentage of Al and Ti in its composition, while Alloy VII is strengthened by a larger addition of refractory metals. The fact that Alloys VII and XIII have idential Al and Ti levels and approximately the same total refractory content, but different strengths, indicates the greater effectiveness of W over Mo as an alloying element. Figure 19 illustrates the way in which these compositional variables manifest themselves in the microstructure of the alloys. Alloy V, Figure 19A, having higher Al and Ti levels but only 11 percent total refractory metal addition, exhibits the eutectic gamma prime structure. Alloys VII and XIII, containing less Al but greater amounts of refractory metals, do not display massive gamma prime formations, Figures 19B and 19C. Grain size is also apparently smaller in these alloys. Figure 19D shows the structure of Alloy XII at a higher magnification. The fine gamma prime dispersion in this alloy can be compared with that in Alloy II, Figure 18B which contains much smaller proportions of the refractory metals.

In all alloys containing greater than 14.5 percent refractory metal additions and 1.8%Ti, an unusual etching phase was encountered as shown in Figure 20. This figure depicts the microstructure observed in Alloy IX which is typical of this group. The structure is shown at 250X (Figures 20A and B) and at 750X magnification (Figure 20C and D) in both the unetched and etched conditions. Comparison of etched and unetched micrographs indicates that the dark areas are etching effects and not voids or inclusions. The proximity of these regions to the "white etching" gamma prime phase which juts into the darker portions, Figure 20D, and the fact that they generally occupy grain boundary areas, leads to the possible conclusion that they are areas of gamma prime intermetallic of a slightly different composition from the lighter areas.

Photomicrographs typical of those alloys containing higher percentages of alloying additions are shown in Figure 21. Figure 21A shows the microstructure of Alloy XVII which contains 4.5Mo, 10.0W, and 1.0Ta in a 6.3Al-1.0Ti base. Along with carbide formations and gamma prime precipitate within the grains, there is a needle-like phase which is possibly a sigma type phase. This phase generally has an embrittling effect upon nickel-base superalloys. Small amounts of this phase were found to appear only in alloys containing more than 13 percent total molybdenum and tungsten, particularly where tantalum was low.

Figures 21B, C, and D show the microstructures ofthe group of alloys containing the largest proportion of refractory metals, Alloys XXV, XXVI, and XXVII. They illustrate the multiplicity of undesirable phases formed when too great an amount of additions are present. Stress rupture results indicate that such an "over-loading" of an alloy is detrimental to high temperature properties. Cross-like patterns are evident in Figure 21C. This structure is particularly characteristic of alloys containing fairly large quantities of tantalum. The effects of additional amounts of aluminum and titanium can also be discerned by comparing Figures 21B, C, and D.

b. <u>Stress Rupture Specimen Evaluation</u>

In analyzing stress rupture results, the actual test data are of the most consequence. However, it is also of significance to examine the nature of the fracture in broken specimens. Both ductile and brittle fractures are encountered. Figure 22 illustrates the typical types of fracture encountered. Figure 23 emphasizes at 60X the flow pattern in the area of the fracture and the intergranular nature of the rupture at 2000°F. The very ductile nature of the fracture in a relatively lean alloy is shown in Figure 22A and 23A. This sample was elongated 73.1 percent and broke upon loading. The pattern of flowed metal can be seen in Figure 23A. The more brittle nature of the fracture in alloys providing superior stress rupture life is shown in Figures 22B and C and Figure 23B. Alloys V and VII which elongated 9.7 percent and 6.6 percent respectively are shown. The intergranular nature of the fracture is especially evident in Figure 23B. Secondary cracking, which indicates brittle behavior, but often adds falsely to the elongation is shown in Figure 22D, in Alloy XXI. Figure 23D illustrates the microstructure at 60X magnification of Alloy XXI which elongated 4.2 percent and had a life of only 2.0 hours.

4. **<u>Workability</u>**

Workability in a nickel-base alloy is important if it is desired to produce the alloy as a wrought material. In this case, the metal must be fabricated from a cast ingot to a finished forging. In order to gage the potential adaptability of the primarily cast compositions developed in this evaluation to wrought procedures, it is desirable to subject the alloy to a workability test. The proposed test, which consists of extrusion to bar stock and side pressing, has been discussed previously. The down pole sections of the cast clusters have been clad in mild steel and extruded at 2075°F to a 4:1 area reduction. Typical extrusions are shown in Figure 24. While the steel canning material has not been removed, these extrusions appear to be sound. There does not appear to be any difference in product quality nor were there any variations in extrusion parameters necessary as a result of the wide range of compositions tested. The extruded bars will next be machined to sound superalloy bars approximately one inch in length and deformed by flat pressing 50 percent between flat dies at various temperatures from 2000°F to 2200°F.

B. Series II and III Alloys (9 compositions each)

Since mechanical property testing of the Series II and III alloys (see Table VII) is just beginning at this reporting period (October 1), there are very few results to be reported. Qualitative observations of the casting procedure indicated that these alloys possess the same excellent castability as did the Series I alloys. Metal additions (including hafnium) were made easily, without observable loss. The precious metals appeared to be melted into the molten bath without difficulty. Again, non-metallic contaminants were minimal. X-ray inspection of the tensile bar preforms indicated 100 percent radiographically sound castings among all 18 heats of Series II and III. A portion of the machined bars have been fluorescent dye penetrant inspected with no indications of surface voids, grinding checks, or other imperfections. It may thus be concluded that the Series II and III alloys possess adequate castability and machinability for use as a precision cast superalloy.

V CONCLUSIONS AND FUTURE WORK

Through the use of a statistical treatment, it has been possible to evaluate the effects of molybdenum, tungsten, and tantalum variations in three nickel-base superalloy systems, each of which contained different Al plus Ti levels. Tantalum was the most effective strengthener with tungsten second. Moderate amounts (18-19 percent) of total refractory metal addition were most beneficial when the aluminum and titanium levels were low (4.5 percent and 1.0 percent respectively). When the aluminum level was increased to 6.3 percent, the greatest strength was derived when refractory content was lower (about 11 percent). Titanium levels above 1.0 percent were generally detrimental. Metallographic studies indicated the presence of several intermetallic phases and morphologies. Alloys having little intermetallic formation had low strength while those possessing gross formations were often brittle.

Based on the statistical and metallographic results of Series I alloy investigations, the following base composition was selected:

Co	Cr	Zr	B	C	Ni	Mo	W	Ta	Ti
10.0	10.0	0.03	0.02	0.13	Bal.	2.0	5.5	8.0	1.0

The effects of Cb, V, Hf, Re, and Ru additions will be evaluated in this base in the Series II and III alloys. The aluminum level will also be varied. After the significance of these elements has been determined, another base composition will be selected which will incorporate all elements deemed beneficial to stress rupture properties. The alloying elements which had previously fixed (i.e., Cr, Co, Zr, B, and C) will then be adjusted for optimum properties.

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TABLE I

Nominal Compositions of Commercial Nickel-Base Superalloys

Alloy	Cast Or Wrought	Cr	Co	Fe	Mo	W	Ti	Al	Cb	C	B	Other
Nimonic 75	W	20	-	-	-	-	-	-	-	0.04	-	
Inconel X	W	15	-	7	-	-	2.5	0.9	0.9	.04	-	
Waspaloy	W	19	14	-	4.3	-	3.0	1.3	-	.06	.005	
Rene' 41	W	19	10	-	10	-	3.0	1.5	-	.10	.005	
Udimet 700	W	15	18	-	5	-	3.5	4.3	-	.08	.03	
Inco 713C	C	13	-	-	-	-	0.6	6.0	-	.06	.015	
TRW 1800	C	13	-	-	-	9	0.6	6.0	1.5	.09	.05	
TRW 1900	C	10.3	10	-	-	9	1.0	6.3	1.5	.12	.03	
IN 100	C	9.5	15	-	3	-	5.0	5.3	-	.15	.015	1.0 V
MarM-200	C	9	10	-	-	12	2.0	5.0	1.0	.15	.02	
TAZ 8	C	6	-	-	4	4	-	6.0	-	.12	-	8.0 Ta 1.0 Zr 2.5 V

TABLE II

Stress Rupture Properties of Contemporary Alloys

A. 1800°F Tests

Hours of Life for Given Stress

Alloy	10	100	1000	10,000
	Wrought Nickel-Base Alloys			
Rene' 41	17 (ksi)	9 (ksi)		
Udimet 700	26	16	7 (ski)	
	Cast Nickel-Base Alloys			
Inco 713 C	29	21	13	8 (ksi)
TRW 1800	31	21	14	9
TRW 1900	34	26	18	14
IN 100		25	15	
MarM-200	37	28	20	15
TAZ 8	35	23	15(1815°F)	

B. 1900°F Tests

Hours of Life for Given Stress

Alloy	10	100	1000	10,000
	Wrought Nickel-Base Alloys			
Udimet 700	17 (ksi)	8 (ksi)	3 (ksi)	
	Cast Nickel-Base Alloys			
TRW 1800	14	8		
TRW 1900	25	16	11	8 (ksi)
IN 100		16	9	
MarM-200	23	18	11	8
TAZ 8		15(1915°F)		

TABLE III

Melting Procedure for Virgin Heats (Series I, II, and III)

Charging

1. Charge in Crucible *(Taycor or Magnorite)
   Ni,Co,Cr,W,Mo,Ta,Cb,V,Ru,Re
2. Charge in First Hopper
   C
3. Charge in Second Hopper
   Al
4. Charge in Third Hopper*
   Ti, Zr, NiB, Hf

Melting

1. Pump down to 10μ pressure (maximum)
2. Power on - melt down
3. Heat to 3000°F - hold until bath is quiet
4. Cool to form skin
5. Add carbon on skin - melt in slowly
6. After carbon boil ceases and pressure returns to below 10μ add aluminum - melt and stir five (5) minutes
7. After aluminum is melted - add titanium, zirconium, nickel boride and hafnium
8. Stir under low power five (5) minutes
9. Cool until crystals appear - record temperature
10. Adjust temperature to 2550°F - 2575°F or at least 50°F above freezing temperature and pour
11. Cure under vacuum ten (10) minutes

*Cb, V, Ru, Re, and Hf were only added in Series II and III where required.

For alloys 3b and 3c ruthenium metal was added from a charging hopper in step 6 after the carbon boil.

Rhenium and ruthenium metal powder was wrapped in Al foil and overwrapped with Ni foil prior to charging.

TABLE IV

Procedure For Re-Melting To Cast Tensile Bar Clusters

1. Preheat molds for 4 hours in air at 1700°F.
2. Charge "master" ingot into crucible liner (magnorite).
3. Position the preheated mold on the furnace turn-table.
4. Pump down to 10μ pressure (maximum)*
5. Melt and heat to 2900°F.
6. Cool to 2825°F and pour into cluster molds.
7. Cure 15 minutes under vacuum.
8. Remove from investment and mold after 2 hours.

*For Series I alloys power was turned on at the start of pump down.

TABLE V

Pertinent Melting Data for Series I, II, and III Heats

	Alloy	Virgin Heats Total Time	Virgin Heats "Silverfish" Temperature	Re-Melt Heats Total Time	Re-Melt Heats Pressure at 2900°F
Series I					
	I	1:16	2550°F	0:16	7.0
	II	1:12	2520	0:23	5.8
	III	1:05	2475	0:12	6.5
	IV	1:00	2500	0:12	6.5
	V	0:54	2480	0:12	7.5
	VI	1:06	2450	0:11	5.5
	VII	1:26	2460	0:10	6.0
	VIII	1:15	2450	0:14	7.5
	IX	0:57	2550	0:13	6.5
	X	1:07	2470	0:12	6.2
	XI	0:53	2500	0:10	7.0
	XII	0:58	2470	0:17	3.5
	XIII	1:06	2480	0:12	7.0
	XIV	0:52	2490	0:12	6.8
	XV	1:10	2400	0:14	5.0
	XVI	1:08	2480	0:14	6.0
	XVII	1:07	2440	0:16	6.0
	XVIII	N.A.	2420	0:11	7.0
	XIX	1:06	2440	0:10	7.0
	XX	0:59	2420	0:14	7.0
	XXI	1:00	2500	0:12	7.0
	XXII	0:46	2450	0:16	N.A.
	XXIII	1:12	2460	0:14	7.0
	XXIV	1:10	2440	0:12	6.0
	XXV	0:54	2430	0:11	6.0
	XXVI	0:55	2480	0:12	6.0
	XXVII	1:15	2490	0:12	7.0
Series II					
	2a	1:17	2450	0:26	8.0
	2b	1:12	2490	0:20	5.0
	2c	1:01	2440	0:21	5.0
	2d	1:13	2460	0:22	3.5
	2e	1:00	2470	0:21	4.5
	2f	1:12	2430	0:23	4.0
	2g	1:26	2425	0:26	3.0
	2h	1:05	2430	0:25	4.0
	2i	1:18	2425	0:18	4.0

TABLE V (continued)

Pertinent Melting Data for Series I, II, and III Heats

	Alloy	Virgin Heats Total Time	Virgin Heats "Silverfish" Temperature	Re-Melt Heats Total Time	Re-Melt Heats Pressure at 2900°F
Series III					
	3a	1:18	2430	0:22	5.0
	3b	1:12	2460	0:22	4.0
	3c	1:06	2475	0:21	4.0
	3d	1:13	2500	0:21	4.4
	3e	0:57	2480	0:22	4.0
	3f	0:52	2480	0:21	3.5
	3g	1:11	2450	0:21	6.5
	3h	0:49	2520	0:20	3.5
	3i	0:53	2510	0:26	4.0

TABLE VI

Aim and Actual Chemical Composition for Series I Alloys
(Aim is shown first)

Alloy	Co	Cr	Zr	B	C	Ni	Mo	W	Ta	Al	Ti	Si	Fe	S	Mn
I	10.0	10.0	0.03	0.02	0.13	Bal	1.0	1.0	1.0	4.5	1.0	-	-	-	-
	10.20	9.79	0.07	0.017	0.09	-	1.04	1.14	1.05	3.99	1.13	0.039			
II	10.0	10.0	0.03	0.02	0.13	Bal	1.0	1.0	1.0	6.3	1.0	-	-	-	-
	10.15	9.70	0.06	0.019	0.10	-	1.03	1.02	1.00	6.45	1.09	0.026			
III	10.0	10.0	0.03	0.02	0.13	Bal	1.0	1.0	1.0	6.3	1.8	-	-	-	-
	10.20	9.86	0.05	0.024	0.11	-	1.05	1.05	1.07	5.93	1.89	0.034			
IV	10.0	10.0	0.03	0.02	0.13	Bal	1.0	5.5	4.5	4.5	1.0	-	-	-	-
	10.10	9.64	0.04	0.017	0.12	-	1.07	5.52	4.69	4.72	1.03	0.062		0.005	
V	10.0	10.0	0.03	0.02	0.13	Bal	1.0	5.5	4.5	6.3	1.0	-	-	-	-
	10.16	9.76	0.05	0.030	0.13	-	1.07	5.18	4.54	6.41	1.02				
VI	10.0	10.0	0.03	0.02	0.13	Bal	1.0	5.5	4.5	6.3	1.8	-	-	-	-
	10.18	9.72	0.05	0.018	0.11	-	1.05	5.67	4.40	6.37	1.86	0.053			
VII	10.0	10.0	0.03	0.02	0.13	Bal	1.0	10.0	8.0	4.5	1.0	-	-	-	-
	10.10	9.80	0.05	0.015	0.13	-	1.03	10.06	7.95	4.50	1.02	0.034			
VIII	10.0	10.0	0.03	0.02	0.13	Bal	1.0	10.0	8.0	6.3	1.0	-	-	-	-
	10.15	9.71	0.03	0.015	0.14	-	1.05	9.95	7.90	6.29	1.03	0.050			
IX	10.0	10.0	0.03	0.02	0.13	Bal	1.0	10.8	8.0	6.3	1.3	-	-	-	-
	10.10	9.68	0.04	0.019	0.13	-	1.11	9.84	7.95	6.34	1.81	0.028			
X	10.0	10.0	0.03	0.02	0.13	Bal	4.5	1.0	4.5	4.5	1.0	-	-	-	-
	10.00	9.76	0.02	0.017	0.13	-	4.55	1.08	4.27	4.52	1.05	0.015			

TABLE VI (continued)

Alloy	Co	Cr	Zr	B	C	Ni	Mo	W	Ta	Al	Ti	Si	Fe	S	Mn
XI	10.0	10.0	0.03	0.02	0.13	Bal	4.5	1.0	4.5	6.3	1.0	-	-	-	-
	10.00	9.79	0.04	0.021	0.13	-	4.51	1.01	4.63	6.27	1.00	0.050			
XII	10.0	10.0	0.03	0.02	0.13	Bal	4.5	1.0	4.5	6.3	1.8	-	-	-	-
	10.14	9.94	0.06	0.021	0.13	-	4.28	0.66	4.29	4.60	1.89				
XIII	10.0	10.0	0.03	0.02	0.13	Bal	4.5	5.5	8.0	4.5	1.0	-	-	-	-
	10.20	9.68	0.02	0.017	0.12	-	4.61	5.63	7.91	4.79	1.03	0.028			
XIV	10.0	10.0	0.03	0.02	0.13	Bal	4.5	5.5	8.0	6.3	1.0	-	-	-	-
	10.08	9.65	0.03	0.020	0.12	-	4.46	5.21	7.74	6.37	1.03				
XV	10.0	10.0	0.03	0.02	0.13	Bal	4.5	5.5	8.0	6.3	1.8	-	-	-	-
	10.00	9.70	0.05	0.017	0.12	-	4.50	5.58	7.82	6.32	1.91	0.034	0.07	0.005	<0.005
XVI	10.0	10.0	0.03	0.02	0.13	Bal	4.5	10.0	1.0	4.5	1.0	-	-	-	-
	10.13	9.78	0.04	0.018	0.13	-	4.55	9.68	1.05	4.50	1.03				
XVII	10.0	10.0	0.03	0.02	0.13	Bal	4.5	10.0	1.0	6.3	1.0	-	-	-	-
	10.11	9.61	0.05	0.017	0.12	-	4.61	10.45	1.21	6.45	1.09	0.038		0.006	
XVIII	10.0	10.0	0.03	0.02	0.13	Bal	4.5	10.0	1.0	6.3	1.8	-	-	-	-
	10.02	9.75	0.05	0.017	0.13	-	4.58	9.95	1.06	6.05	1.86	0.047			
XIX	10.0	10.0	0.03	0.02	0.13	Bal	8.0	1.0	8.0	4.5	1.0	-	-	-	-
	10.05	9.71	0.04	0.015	0.13	-	7.92	1.09	7.92	4.54	1.05	0.046			
XX	10.0	10.0	0.03	0.02	0.13	Bal	8.0	1.0	8.0	6.3	1.0	-	-	-	-
	10.13	9.79	0.05	0.019	0.12	-	7.89	1.21	7.49	6.24	1.05				

TABLE VI (continued)

Alloy	Co	Cr	Zr	B	C	Ni	Mo	W	Ta	Al	Ti	Si	Fe	S	Mn
XXI	10.0	10.0	0.03	0.02	0.13	Bal	8.0	1.0	8.0	6.3	1.8	-	-	-	-
	10.18	9.69	0.05	0.021	0.13	-	7.98	0.76	7.73	6.27	1.92				
XXII	10.0	10.0	0.03	0.02	0.13	Bal	8.0	5.5	1.0	4.5	1.0	-	-	-	-
	10.06	9.83	0.04	0.020	0.13	-	7.96	5.54	1.06	4.52	1.05				
XXIII	10.0	10.0	0.03	0.02	0.13	Bal	8.0	5.5	1.0	6.3	1.0	-	-	-	-
	10.12	9.60	0.03	0.021	0.12	-	7.98	5.35	1.14	6.20	1.06				
XXIV	10.0	10.0	0.03	0.02	0.13	Bal	8.0	5.5	1.0	6.3	1.8	-	-	-	-
	10.06	9.53	0.06	0.021	0.10	-	7.98	5.17	1.06	6.47	1.92				
XXV	10.0	10.0	0.03	0.02	0.13	Bal	8.0	10.0	4.5	4.5	1.0	-	-	-	-
	10.18	9.83	0.04	0.023	0.13	-	7.92	10.20	4.34	4.51	1.05			0.005	
XXVI	10.0	10.0	0.03	0.02	0.13	Bal	8.0	10.0	4.5	6.3	1.0	-	-	-	-
	10.01	9.78	0.03	0.020	0.13	-	7.81	10.10	4.38	6.33	1.03				
XXVII	10.0	10.0	0.03	0.02	0.13	Bal	8.0	10.0	4.5	6.3	1.8				
	10.14	9.71	0.04	0.021	0.13	-	7.84	10.26	4.26	6.53	1.98				

TABLE VII

Chemical Composition of Series II and III Alloys (Aim Analysis)

Alloy	Co	Cr	Zr	B	C	Ni	Mo	W	Ta	Ti	Al	Ru	Re	Cb	V	Hf
2a	10.0	10.0	0.03	0.02	0.13	Bal	2.0	5.5	8.0	1.0	4.5	-	-	0	0	0
2b	10.0	10.0	0.03	0.02	0.13	Bal	2.0	5.5	8.0	1.0	4.5	-	-	0	1.0	1.0
2c	10.0	10.0	0.03	0.02	0.13	Bal	2.0	5.5	8.0	1.0	4.5	-	-	0	2.0	2.0
2d	10.0	10.0	0.03	0.02	0.13	Bal	2.0	5.5	8.0	1.0	4.5	-	-	1.0	0	1.0
2e	10.0	10.0	0.03	0.02	0.13	Bal	2.0	5.5	8.0	1.0	4.5	-	-	1.0	1.0	2.0
2f	10.0	10.0	0.03	0.02	0.13	Bal	2.0	5.5	8.0	1.0	4.5	-	-	1.0	2.0	0
2g	10.0	10.0	0.03	0.02	0.13	Bal	2.0	5.5	8.0	1.0	4.5	-	-	2.0	0	2.0
2h	10.0	10.0	0.03	0.02	0.13	Bal	2.0	5.5	8.0	1.0	4.5	-	-	2.0	1.0	0
2i	10.0	10.0	0.03	0.02	0.13	Bal	2.0	5.5	8.0	1.0	4.5	-	-	2.0	2.0	1.0
3a	10.0	10.0	0.03	0.02	0.13	Bal	2.0	5.5	8.0	1.0	6.3	0	0	-	-	-
3b	10.0	10.0	0.03	0.02	0.13	Bal	2.0	5.5	8.0	1.0	5.4	1.0	0	-	-	-
3c	10.0	10.0	0.03	0.02	0.13	Bal	2.0	5.5	8.0	1.0	4.5	2.0	0	-	-	-
3d	10.0	10.0	0.03	0.02	0.13	Bal	2.0	5.5	8.0	1.0	5.4	0	2.0	-	-	-
3e	10.0	10.0	0.03	0.02	0.13	Bal	2.0	5.5	8.0	1.0	4.5	1.0	2.0	-	-	-
3f	10.0	10.0	0.03	0.02	0.13	Bal	2.0	5.5	8.0	1.0	6.3	2.0	2.0	-	-	-
3g	10.0	10.0	0.03	0.02	0.13	Bal	2.0	5.5	8.0	1.0	4.5	0	4.0	-	-	-
3h	10.0	10.0	0.03	0.02	0.13	Bal	2.0	5.5	8.0	1.0	6.3	1.0	4.0	-	-	-
3i	10.0	10.0	0.03	0.02	0.13	Bal	2.0	5.5	8.0	1.0	5.4	2.0	4.0	-	-	-

TABLE VIII

Stress Rupture Data for Series I Alloys Tested at 2000°F 15,000 psi

Alloy	Compositional Aims (Wt. %) Mo	W	Ta	Al	Ti	Life (hrs.)	$\log_{10}$ (Lifex10^2)	Elong. (%)	R.A. (%)
I	1.0	1.0	1.0	4.5	1.0	0.01*	0.0000	73.1	92.4
						0.01*	0.0000	72.6	95.6
						0.01*	0.0000	73.5	86.2
II	1.0	1.0	1.0	6.3	1.0	0.07	0.8451	41.4	42.2
						0.06	0.7782	40.3	34.6
						0.08	0.9031	38.6	32.8
III	1.0	1.0	1.0	6.3	1.8	0.10	1.000	30.2	35.0
						0.13	1.1139	28.0	35.0
						0.10	1.0000	44.7	36.4
IV	1.0	5.5	4.5	4.5	1.0	0.24	1.3802	23.3	24.1
						0.30	1.4771	23.8	25.0
						0.21	1.3222	28.8	35.8
V	1.0	5.5	4.5	6.3	1.0	9.4	2.9736	9.7	9.0
						12.4	3.0934	7.9	7.0
						13.8	3.1399	8.1	8.5
VI	1.0	5.5	4.5	6.3	1.8	5.6	2.7482	3.5	2.0
						12.9	3.1106	3.3	1.2
						5.8	2.7634	1.8	2.0
						5.4**		1.1	2.0
						9.2**		0.8	1.2
VII	1.0	10.0	8.0	4.5	1.0	14.0	3.1461	6.6	13.0
						16.7	3.2227	2.6	7.0
						18.5	3.2672	7.4	8.1

*Broke on loading. Assigned arbitrary value of 0.01 hrs.

**Additional tests were conducted in an attempt to verify previous results. These data were not used in the Latin Square calculations.

TABLE VIII (continued)

Stress Rupture Data for Series I Alloys Tested at 2000°F 15,000 psi

Alloy	Mo	W	Ta	Al	Ti	Life (hrs.)	Log_{10} ($Lifex10^2$)	Elong. (%)	R.A. (%)
		Compositional Aims (Wt. %)							
XVI	4.5	10.0	1.0	4.5	1.0	0.9	1.9542	26.1	39.4
						1.1	2.0414	25.9	38.1
						1.0	2.0000	25.9	47.9
XVII	4.5	10.0	1.0	6.3	1.0	2.2	2.3424	6.7	8.6
						2.2	2.3424	5.9	7.7
						2.8	2.4472	7.4	8.6
XVIII	4.5	10.0	1.0	6.3	1.8	5.3	2.7243	7.0	4.0
						7.8	2.8921	4.5	8.2
						4.9	2.6902	7.6	7.0
XIX	8.0	1.0	8.0	4.5	1.0	5.3	2.7243	5.3	7.4
						6.1	2.7853	3.7	5.1
						5.7	2.7559	6.9	5.5
XX	8.0	1.0	8.0	6.3	1.0	2.5	2.3979	2.7	4.5
						2.3	2.3617	4.1	4.7
						2.8	2.4472	5.1	5.1
XXI	8.0	1.0	8.0	6.3	1.8	2.0	2.3010	4.2	4.3
						2.5	2.3979	4.7	4.9
						2.3	2.3617	4.2	3.9
XXII	8.0	5.5	1.0	4.5	1.0	0.21	1.3222	25.5	54.9
						0.32	1.5051	40.3	61.5
						0.24	1.3802	29.2	62.8
XXIII	8.0	5.5	1.0	6.3	1.0	1.3	2.1139	9.1	14.9
						1.5	2.1761	9.0	12.0
						1.2	2.0792	10.3	15.2
XXIV	8.0	5.5	1.0	6.3	1.0	2.7	2.4314	6.4	10.1
						2.6	2.4150	5.6	7.8
						2.6	2.4150	7.1	7.8

TABLE VIII (continued)

Stress Rupture Data for Series I Alloys Tested at 2000°F 15,000 psi

Alloy	Compositional Aims (Wt. %) Mo	W	Ta	Al	Ti	Life (hrs.)	Log_{10} (Lifex10^2)	Elong. (%)	R.A. (%)
VIII	1.0	10.0	8.0	6.3	1.0	5.4	2.7324	2.5	3.1
						2.2	2.3424	3.3	3.9
						6.2	2.7924	2.8	3.6
IX	1.0	10.0	8.0	6.3	1.8	2.0	2.3010	3.4	3.1
						2.6	2.4150	7.0	2.7
						0.6	1.7782	6.5	4.7
X	4.5	1.0	4.5	4.5	1.0	0.30	1.4771	29.8	36.2
						0.24	1.3802	23.1	37.5
						0.18	1.2553	33.1	31.6
XI	4.5	1.0	4.5	6.3	1.0	4.7	2.6721	1.4	3.5
						3.9	2.5911	4.2	7.7
						3.7	2.5682	7.6	8.5
XII	4.5	1.0	4.5	6.3	1.8	2.1	2.3222	16.9	22.4
						2.5	2.3979	11.8	15.5
						2.7	2.4314	17.2	23.1
XIII	4.5	5.5	8.0	4.5	1.0	8.3	2.9191	4.0	6.6
						9.5	2.9777	6.1	6.2
						7.6	2.8808	4.9	7.4
XIV	4.5	5.5	8.0	6.3	1.0	4.7	2.6721	4.4	2.0
						4.2	2.6232	2.3	2.4
						3.5	2.5441	2.5	2.8
XV	4.5	5.5	8.0	6.3	1.8	3.0	2.4771	4.7	3.2
						2.2	2.3424	3.8	3.6
						2.3	2.3617	5.4	4.7

TABLE VIII (continued)

Stress Rupture Data for Series I Alloys Tested at 2000°F 15,000 psi

	Compositional Aims (Wt. %)								
Alloy	Mo	W	Ta	Al	Ti	Life (hours.)	Log_{10} ($Life x 10^2$)	Elong. (%)	R.A. (%)
XXV	8.0	10.0	4.5	4.5	1.0	1.4	2.1461	9.4	14.6
						1.5	2.1761	9.3	16.2
						1.8	2.2553	12.2	15.7
XXVI	8.0	10.0	4.5	6.3	1.0	0.35	1.5441	14.9	19.2
						0.35	1.5441	16.3	23.1
						0.42	1.6232	14.8	20.5
XXVII	8.0	10.0	4.5	6.3	1.8	0.25	1.3979	12.6	17.4
						0.24	1.3802	10.4	13.5
						0.20	1.3010	14.6	21.0

TABLE IX

2000°F Stress Rupture Life

(Life in Hours Converted to Log_{10} (Hours x 10^2) For Computation)

A. 4.5% Al - 1.0 Ti Base

Tungsten	Molybdenum 1.0		4.5		8.0	
1.0	0.0000 0.0000 0.0000	(1.0)*	1.4771 1.3802 1.2553	(4.5)	2.7243 2.7853 2.7559	(8.0)
5.5	1.3802 1.4771 1.3222	(4.5)	2.9191 2.9777 2.8808	(8.0)	1,3222 1.5051 1.3802	(1.0)
10.0	3.1461 3.2227 3.2672	(8.0)	1.9542 2.0414 2.0000	(1.0)	2.1461 2.1761 2.2553	(4.5)

* Numbers in Parentheses are Tantalum Contents.

Average Test Value

	1%	4.5% or 5.5%	8.0% or 10.0%
Mo	1.5351 (0.34)**	2.0984 (1.25)	2.1167 (1.31)
W	1.3753 (0.24)	1.9072 (0.81)	2.4677 (2.94)
Ta	1.1337 (0.14)	1.6522 (0.45)	2.9643 (9.21)

** Average Life (Hours) is Given in Parenthesis

	Sum of Squares	Degrees of Freedom	Mean Square	F Ratio***
Mo	1.9682	2	0.9841	28.52
W	5.3706	2	2.6853	77.83
Ta	16.0258	2	8.0129	232.26
Residual and Interaction	0.6907	20	0.0345	
Total	24.0553	26		

*** F Ratio for 99% Significance Level $\geq$ 5.85

TABLE IX (continued)

2000°F Stress Rupture Life

(Life in Hours Converted to Log_{10} (Hours x 10^2) For Computation)

B. 6.3% Al - 1.0 Ti Base

Tungsten	Molybdenum 1.0		Molybdenum 4.5		Molybdenum 8.0	
1.0	0.8451 0.7782 0.9031	(1.0)*	2.6721 2.5911 2.5682	(4.5)	2.3979 2.3617 2.4472	(8.0)
5.5	2.9736 3.0934 3.1399	(4.5)	2.6721 2.6232 2.5441	(8.0)	2.1139 2.1761 2.0792	(1.0)
10.0	2.7324 2.3424 2.7924	(8.0)	2.3424 2.3424 2.4472	(1.0)	1.5441 1.5441 1.6232	(4.5)

* Numbers in Parentheses are Tantalum Contents

Average Test Values

	1%	4.5% or 5.5%	8.0% or 10.0%
Mo	2.1778 (1.51)**	2.5336 (3.42)	2.0319 (1.08)
W	1.9516 (0.90)	2.6017 (4.00)	2.1901 (1.55)
Ta	1.7808 (0.60)	2.4166 (2.61)	2.5459 (3.52)

** Average Life (Hours) is Given in Parenthesis

	Sum of Squares	Degrees of Freedom	Mean Square	F Ratio***
Mo	1.1988	2	0.5994	2.83
W	1.9468	2	0.9734	4.59
Ta	3.5023	2	1.7512	8.26
Residual and Interaction	4.2399	20	0.2120	
Total	10.8878	26		

*** F Ratio for 99% Significance Level $\geq$ 5.85

TABLE IX (continued)

2000°F Stress Rupture Life

(Life in Hours Converted to Log_{10} (Hours x 10^2) for Computation)

C. 6.3% Al - 1.8 Ti Base

Tungsten	Molybdenum 1.0		Molybdenum 4.5		Molybdenum 8.0	
1.0	1.0000		2.3222		2.3010	
	1.1139	(1.0)*	2.3979	(4.5)	2.3979	(8.0)
	1.0000		2.4314		2.3617	
5.5	2.7482		2.4771		2.4314	
	3.1106	(4.5)	2.3424		2.4150	(1.0)
	2.7634		2.3617		2.4150	
10.0	2.3010		2.7243		1.3979	
	2.4150	(8.0)	2.8921	(1.0)	1.3802	(4.5)
	1.7782		2.6902		1.3010	

* Numbers in Parentheses are Tantalum Contents.

Average Test Values

	1%		4.5% or 5.5%		8.0% or 10.0%	
Mo	2.0256	(1.06)**	2.5155	(3.28)	2.0446	(1.11)
W	1.9251	(0.84)	2.5628	(3.65)	2.0978	(1.25)
Ta	2.0758	(1.19)	2.2059	(1.61)	2.3040	(2.01)

** Average Life (Hours) is Given in Parenthesis.

	Sum of Squares	Degrees of Freedom	Mean Square	F Ratio***
Mo	1.3864	2	0.6932	2.42
W	1.9578	2	0.9789	3.42
Ta	0.2359	2	0.1180	0.41
Residual and Interaction	5.7306	20	0.2865	
Total	9.3107	26		

*** F Ratio for 99% Significance Level $\geq$ 5.85

TABLE X

2000°F Stress Rupture Elongation (%)

A. 4.5% Al - 1.0 Ti Base

Tungsten	Molybdenum 1.0		Molybdenum 4.5		Molybdenum 8.0	
1.0	73.1 72.6 73.5	(1.0)*	29.8 23.1 33.1	(4.5)	5.3 3.7 6.9	(8.0)
5.5	23.3 23.8 28.8	(4.5)	4.0 6.1 4.9	(8.0)	25.5 40.3 29.2	(1.0)
10.0	6.6 2.6 7.4	(8.0)	26.1 25.9 25.9	(1.0)	9.4 9.3 12.2	(4.5)

*Numbers in Parentheses are Tantalum Contents.

Average Test Values

	1%	4.5% or 5.5%	8.0% or 10.0%
Mo	34.6	19.9	15.8
W	35.7	20.7	13.9
Ta	43.6	21.4	5.3

	Sum of Squares	Degrees of Freedom	Mean Square	F Ratio**
Mo	1773.27	2	886.64	23.64
W	2231.03	2	1115.52	29.74
Ta	6651.17	2	3325.59	88.66
Residual and Interaction	750.16	20	37.51	
Total	11405.63	26		

**F Ratio for 99% Significance Level ≥ 5.85

TABLE X (continued)

2000°F Stress Rupture Elongation (%)

B. 6.3% Al - 1.0 Ti Base

Tungsten	Molybdenum 1.0		Molybdenum 4.5		Molybdenum 8.0	
1.0	41.4 40.3 38.6	(1.0)*	1.4 4.2 7.6	(4.5)	2.7 4.1 5.1	(8.0)
5.5	9.7 7.9 8.1	(4.5)	4.4 2.3 2.5	(8.0)	9.1 9.0 10.3	(1.0)
10.0	2.5 3.3 2.8	(8.0)	6.7 5.9 7.4	(1.0)	14.9 16.3 14.8	(4.5)

* Numbers in Parentheses are Tantalum Contents

Average Test Values

	1%	4.5% or 5.5%	8.0% or 10.0%
Mo	17.2	4.7	9.6
W	16.2	7.0	8.3
Ta	18.7	9.4	3.3

	Sum of Squares	Degrees of Freedom	Mean Square	F Ratio**
Mo	710.40	2	355.20	6.25
W	440.03	2	220.02	3.87
Ta	1088.54	2	544.27	9.58
Residual and Interaction	1135.85	20	56.79	
Total	3374.82	26		

**F Ratio for 99% Significance Level ≥ 5.85

TABLE X (continued)

2000°F Stress Rupture Elongation (%)

C. 6.3% Al - 1.8 Ti Base

Tungsten	Molybdenum 1.0		4.5		8.0	
1.0	30.2 28.0 44.7	(1.0)*	16.9 11.8 17.2		4.2 4.7 4.2	(8.0)
5.5	3.5 3.3 1.8	(4.5)	4.7 3.8 5.4	(8.0)	6.4 5.6 7.1	(1.0)
10.0	3.4 7.0 6.5	(8.0)	7.0 4.5 7.6	(1.0)	12.6 10.4 14.6	(4.5)

* Numbers in Parentheses are Tantalum Contents.

Average Test Value

	1%	4.5% or 5.5%	8.0% or 10.0%
Mo	14.3	8.8	7.8
W	18.0	4.6	8.2
Ta	15.7	10.2	4.9

	Sum of Squares	Degrees of Freedom	Mean Square	F Ratio**
Mo	221.00	2	110.50	2.34
W	862.69	2	431.35	9.13
Ta	524.88	2	262.44	5.56
Residual and Interaction	944.45	20		
Total	2553.02	26		

** F Ratio for 99% Significance Level $\geq$ 5.85

TABLE XI

2000°F Tensile Results For Series I Alloys

	Compositional Aims (Wt.%)								
Alloy	Mo	W	Ta	Al	Ti	Ultimate, 1000psi	0.2% Offset Yield, 1000psi	Elong (%)	R.A. (%)
I	1.0	1.0	1.0	4.5	1.0	5.0 5.0	4.6 4.4	72.6 71.8	94.4 92.4
II	1.0	1.0	1.0	6.3	1.0	11.1 11.1	9.3 9.3	42.1 32.5	46.3 40.1
III	1.0	1.0	1.0	6.3	1.8	11.3 11.9	9.8 9.9	35.6 42.9	31.1 47.6
IV	1.0	5.5	4.5	4.5	1.0	16.0 15.6	12.8 13.0	30.3 26.4	22.0 23.7
V	1.0	5.5	4.5	6.3	1.0	32.5 31.3	29.1 27.1	17.5 13.2	13.6 10.4
VI	1.0	5.5	4.5	6.3	1.8	34.4 35.5	30.9 29.9	5.2 3.7	5.1 3.1
VII	1.0	10.0	8.0	4.5	1.0	36.1 34.7	30.5 29.9	10.0 15.7	13.3 15.1
VIII	1.0	10.0	8.0	6.3	1.0	32.7 32.6	29.3 29.4	5.5 7.2	8.2 8.9
IX	1.0	10.0	8.0	6.3	1.8	31.6 31.4	28.6 29.5	10.7 6.9	11.8 5.5
X	4.5	1.0	4.5	4.5	1.0	14.3 14.3	11.8 12.0	27.1 28.2	30.1 30.7
XI	4.5	1.0	4.5	6.3	1.0	28.3 28.0	24.5 24.1	6.0 12.0	9.3 8.5

TABLE XI (contimued)

2000°F Tensile Results For Series I Alloys

	Compositional Aims (Wt.%)								
Alloy	Mo	W	Ta	Al	Ti	Ultimate, 1000psi	0.2% Offset Yield, 1000psi	Elong (%)	R.A. (%)
XII	4.5	1.0	4.5	6.3	1.8	24.0 23.6	19.2 19.3	15.4 22.8	26.9 23.2
XIII	4.5	5.5	8.0	4.5	1.0	30.5 29.5	25.9 25.0	11.3 10.5	11.5 8.8
XIV	4.5	5.5	8.0	6.3	1.0	30.6 32.3	27.4 29.0	10.6 8.6	11.5 9.3
XV	4.5	5.5	8.0	6.3	1.8	29.0 28.1	26.2 25.8	5.7 8.3	8.9 9.7
XVI	4.5	10.0	1.0	4.5	1.0	23.0 23.1	18.9 18.7	22.2 16.5	31.4 33.7
XVII	4.5	10.0	1.0	6.3	1.0	26.1 24.7	22.6 21.5	10.7 16.2	15.7 15.6
XVIII	4.5	10.0	1.0	6.3	1.8	29.4 24.3	25.8 21.6	13.2 11.4	14.4 34.6
XIX	8.0	1.0	8.0	4.5	1.0	28.3 30.4	23.5 25.8	13.1 13.8	15.9 15.3
XX	8.0	1.0	8.0	6.3	1.0	30.6 30.5	27.2 26.6	8.1 10.8	13.9 10.4
XXI	8.0	1.0	8.0	6.3	1.8	30.3 31.1	27.4 27.8	8.6 10.7	11.0 8.9
XXII	8.0	5.5	1.0	4.5	1.0	17.8 20.0	14.9 15.1	26.5 20.6	44.1 50.3

TABLE XI (continued)

2000°F Tensile Results For Series I Alloys

Alloy	Compositional Aims (Wt.%) Mo	W	Ta	Al	Ti	Ultimate, 1000psi	0.2% Offset Yield, 1000psi	Elong. (%)	R.A. (%)
XXIII	8.0	5.5	1.0	6.3	1.0	24.8 26.0	23.4 21.5	15.6 11.3	24.4 16.8
XXIV	8.0	5.5	1.0	6.3	1.8	30.2 28.9	27.3 25.5	13.0 12.3	12.9 17.0
XXV	8.0	10.0	4.5	4.5	1.0	26.4 25.2	22.4 21.2	15.5 10.9	24.7 22.6
XXVI	8.0	10.0	4.5	6.3	1.0	21.5 21.4	18.9 18.6	16.4 15.1	22.7 15.6
XXVII	8.0	10.0	4.5	6.3	1.8	19.8 19.8	16.2 17.5	17.3 18.5	19.1 19.6

TABLE XII

1875°F Tensile Results For Series I Alloys

Alloy	Compositional Aims (Wt.%) Mo	W	Ta	Al	Ti	Ultimate, 1000psi	0.2% Offset Yield, 1000psi	Elong. (%)	R.A. (%)
I	1.0	1.0	1.0	4.5	1.0	15.0 15.5	14.0 14.3	31.0 22.7	33.8 34.2
II	1.0	1.0	1.0	6.3	1.0	25.5 25.7	20.0 19.8	- 22.3	19.6 22.3
III	1.0	1.0	1.0	6.3	1.8	26.2 25.7	20.6 21.3	19.5 25.0	16.0 25.9
IV	1.0	5.5	4.5	4.5	1.0	29.5 29.7	23.1 23.7	18.7 21.4	18.0 18.8
V	1.0	5.5	4.5	6.3	1.0	51.7 49.9	40.5 44.1	12.8 12.4	11.2 12.7
VI	1.0	5.5	4.5	6.3	1.8	56.9 55.1	49.0 47.9	3.8 4.5	3.5 4.7
VII	1.0	10.0	8.0	4.5	1.0	54.1 56.0	42.4 47.1	9.9 8.5	11.5 10.4
VIII	1.0	10.0	8.0	6.3	1.0	51.2 50.0	46.9 46.0	6.5 5.1	7.0 6.2
IX	1.0	10.0	8.0	6.3	1.8	46.4 47.2	42.2 44.0	5.1 4.0	6.2 5.5
X	4.5	1.0	4.5	4.5	1.0	29.2 29.5	22.3 23.5	21.3 23.9	28.4 25.6
XI	4.5	1.0	4.5	6.3	1.0	46.5 47.7	40.5 41.8	7.1 7.4	8.5 10.4

TABLE XII (continued)

1875°F Tensile Results For Series I Alloys

	Compositional Aims (Wt. %)								
Alloy	Mo	W	Ta	Al	Ti	Ultimate, 1000psi	0.2% Offset Yield, 1000psi	Elong. (%)	R.A. (%)
XII	4.5	1.0	4.5	6.3	1.8	39.4 39.5	31.2 32.5	16.1 16.5	21.0 18.8
XIII	4.5	5.5	8.0	4.5	1.0	50.1 51.0	43.5 44.4	8.1 5.8	9.3 6.8
XIV	4.5	5.5	8.0	6.3	1.0	49.2 48.5	43.8 43.7	7.1 7.3	10.0 11.2
XV	4.5	5.5	8.0	6.3	1.8	45.3 44.5	41.8 40.4	7.9 7.7	10.4 8.6
XVI	4.5	10.0	1.0	4.5	1.0	42.1 43.9	34.2 37.0	13.5 11.9	16.6 11.1
XVII	4.5	10.0	1.0	6.3	1.0	44.5 45.2	38.8 39.9	7.2 9.7	8.5 15.0
XVIII	4.5	10.0	1.0	6.3	1.8	45.9 47.5	41.6 43.0	9.4 8.9	12.6 13.8
XIX	8.0	1.0	8.0	4.5	1.0	44.3 46.7	36.4 38.6	11.7 8.9	13.3 10.0
XX	8.0	1.0	8.0	6.3	1.0	43.7 46.6	37.7 41.5	7.8 5.9	12.8 12.2
XXI	8.0	1.0	8.0	6.3	1.8	45.5 49.5	41.2 44.6	6.8 7.3	14.2 12.8
XXII	8.0	5.5	1.0	4.5	1.0	36.2 37.0	29.3 31.1	14.8 13.3	26.7 22.3

TABLE XII (continued)

1875°F Tensile Results For Series I Alloys

Alloy	Compositional Aims (Wt. %) Mo	W	Ta	Al	Ti	Ultimate, 1000psi	0.2% Offset Yield, 1000psi	Elong. (%)	R.A. (%)
XXIII	8.0	5.5	1.0	6.3	1.0	43.2	37.5	15.4	19.9
						42.9	38.0	11.9	16.8
XXIV	8.0	5.5	1.0	6.3	1.8	46.3	41.4	11.7	17.3
						43.5	38.1	10.2	14.4
XXV	8.0	10.0	4.5	4.5	1.0	44.3	38.4	15.0	22.5
						41.7	36.0	17.5	20.3
XXVI	8.0	10.0	4.5	6.3	1.0	36.4	31.9	12.4	13.1
						34.8	30.4	12.9	15.3
XXVII	8.0	10.0	4.5	6.3	1.8	34.7	30.6	10.3	14.7
						33.2	29.0	9.2	11.9

TABLE XIII

2000°F Tensile Strength, 1000 psi

A. 4.5% Al - 1.0 Ti Base

Tungsten	Molybdenum 1.0		Molybdenum 4.5		Molybdenum 8.0	
1.0	5.0 5.0	(1.0)*	14.3 14.3	(4.5)	28.3 30.4	(8.0)
5.5	16.0 15.6	(4.5)	30.5 29.5	(8.0)	17.8 20.0	(1.0)
10.0	36.1 34.7	(8.0)	23.0 23.1	(1.0)	26.4 25.2	(4.5)

*Numbers in Parentheses are Tantalum Contents.

Average Test Values

	1%	4.5% or 5.5%	8.0% or 10.0%
Mo	18.7	22.5	24.7
W	16.2	21.6	28.1
Ta	15.7	18.6	31.6

	Sum of Sqaures	Degrees of Freedom	Mean Square	F Ratio**
Mo	110.07	2	55.04	19.52
W	423.81	2	211.91	75.15
Ta	860.95	2	430.48	152.65
Residual and Interaction	30.97	11	2.82	
Total	1425.80	17		

** F Ratio for 99% Significance Level ≥ 7.20

TABLE XIII (continued)

2000°F Tensile Strength, 1000 psi

B. 6.3% Al - 1.0 Ti Base

Tungsten	Molybdenum 1.0		Molybdenum 4.5		Molybdenum 8.0	
1.0	11.1 11.1	(1.0)*	28.3 28.0	(4.5)	30.6 30.5	(8.0)
5.5	32.5 31.3	(4.5)	30.6 32.3	(8.0)	24.8 26.0	(1.0)
10.0	32.7 32.6	(8.0)	26.1 24.7	(1.0)	21.5 21.4	(4.5)

* Numbers in Parentheses are Tantalum Contents

Average Test Values

	1%	4.5% or 5.5%	8.0% or 10.0%
Mo	25.2	28.3	25.8
W	23.3	29.6	26.5
Ta	20.6	27.2	31.6

	Sum of Squares	Degrees of Freedom	Mean Square	F Ratio**
Mo	32.94	2	16.47	0.72
W	119.72	2	59.86	2.62
Ta	362.15	2	181.08	7.92
Residual and Interaction	240.45	11	22.86	
Total	755.26	17		

** F Ratio for 99% Significance Level $\geq$ 7.20

TABLE XIII (continued)

2000°F Tensile Strength, 1000 psi

C. 6.3% Al - 1.8 Ti Base

Tungsten	Molybdenum 1.0		Molybdenum 4.5		Molybdenum 8.0	
1.0	11.3 11.9	(1.0)*	24.0 23.6	(4.5)	30.3 31.1	(8.0)
5.5	34.4 35.5	(4.5)	29.0 28.1	(8.0)	30.2 28.9	(1.0)
10.0	31.6 31.4	(8.0)	29.4 24.3	(1.0)	19.8 19.8	(4.5)

*Numbers in Parentheses are Tantalum Contents

Average Test Values

	1%	4.5% or 5.5%	8.0% or 10.0%
Mo	26.0	26.4	26.7
W	22.0	31.0	26.1
Ta	22.7	26.2	30.3

	Sum of Squares	Degrees of Freedom	Mean Square	F Ratio**
Mo	1.34	2	0.67	0.02
W	243.01	2	121.51	3.33
Ta	172.83	2	86.42	2.37
Residual and Interaction	401.68	11	36.52	
Total	818.86	17		

**F Ratio for 99% Significance Level ≥ 7.20

TABLE XIV

1875°F Tensile Strength, 1000 psi

A. 4.5% Al - 1.0 Ti Base

Tungsten	Molybdenum 1.0		Molybdenum 4.5		Molybdenum 8.0	
1.0	15.0 15.5	(1.0)*	29.2 29.5	(4.5)	44.3 46.7	(8.0)
5.5	29.5 29.7	(4.5)	50.1 51.0	(8.0)	36.2 37.0	(1.0)
10.0	54.1 56.0	(8.0)	42.1 43.9	(1.0)	44.3 41.7	(4.5)

*Numbers in Parentheses are Tantalum Contents

Average Test Values

	1%	4.5% or 5.5%	8.0% or 10.0%
Mo	33.3	41.0	41.7
W	30.0	38.9	47.0
Ta	31.6	34.0	50.4

	Sum of Squares	Degrees of Freedom	Mean Square	F Ratio**
Mo	259.75	2	129.88	22.05
W	865.91	2	432.96	73.51
Ta	1251.15	2	625.58	106.21
Residual and Interaction	64.77	11	5.89	
Total	2441.58	17		

** F Ratio for 99% Significance Level $\geq$ 7.20

TABLE XIV (continued)

1875° Tensile Strength, 1000 psi

B. 6.3% Al - 1.0 Ti Base

Tungsten	Molybdenum 1.0		4.5		8.0	
1.0	25.5 25.7	(1.0)*	46.5 47.7	(4.5)	43.7 46.6	(8.0)
5.5	51.7 49.9	(4.5)	49.2 48.5	(8.0)	43.2 42.9	(1.0)
10.0	51.2 50.0	(8.0)	44.5 45.2	(1.0)	36.4 34.8	(4.5)

*Numbers in Parentheses are Tantalum Contents

Average Test Values

	1%	4.5% or 5.5%	8.0% or 10.0%
Mo	42.3	46.9	41.3
W	39.3	47.6	43.7
Ta	37.8	44.5	48.2

	Sum of Squares	Degrees of Freedom	Mean Square	F Ratio**
Mo	108.83	2	54.42	1.40
W	206.11	2	103.06	2.64
Ta	331.21	2	165.61	4.25
Residual and Interaction	428.65	11	38.97	
Total	1074.80	17		

** F Ratio for 99% Significance Level ≥ 7.20

TABLE XIV (continued)

1875°F Tensile Strength, 1000 psi

C. 6.3% Al - 1.8 Ti Base

Tungsten	Molybdenum 1.0		Molybdenum 4.5	Molybdenum 8.0
1.0	26.2 25.7	(1.0)*	39.4 (4.5) 39.5	45.5 (8.0) 49.5
5.5	56.9 55.1	(4.5)	45.3 (8.0) 44.5	46.3 (1.0) 43.5
10.0	46.4 47.2	(8.0)	45.9 (1.0) 47.5	34.7 (4.5) 33.2

* Numbers in Parentheses are Tantalum Contents

Average Test Values

	1%	4.5% or 5.5%	8.0% or 10.0%
Mo	42.9	43.7	42.1
W	37.6	48.6	42.5
Ta	39.2	43.1	46.4

	Sum of Squares	Degrees of Freedom	Mean Square	F Ratio**
Mo	7.36	2	3.68	0.06
W	362.41	2	181.21	2.81
Ta	156.71	2	78.36	1.21
Residual and Interaction	709.69	11	64.52	
Total	1236.17	17		

**F Ratio for 99% Significance Level $\geq$ 7.20

TABLE XV

2000°F Tensile Elongation(%)

A. 4.5% Al - 1.0 Ti Base

Tungsten	Molybdenum 1.0		4.5		8.0	
1.0	72.6 71.8	(1.0)*	27.1 28.2	(4.5)	13.1 13.8	(8.0)
5.5	30.3 26.4	(4.5)	11.3 10.5	(8.0)	26.5 20.6	(1.0)
10.0	10.0 15.7	(8.0)	22.2 16.5	(1.0)	15.5 10.9	(4.5)

* Numbers in Parentheses are Tantalum Contents

Average Test Values

	1%	4.5% or 5.5%	8.0% or 10.0%
Mo	37.8	19.3	16.7
W	37.8	20.9	15.1
Ta	38.4	23.1	12.4

	Sum of Squares	Degrees of Freedom	Mean Square	F Ratio**
Mo	1585.29	2	792.65	15.11
W	1658.55	2	829.28	15.81
Ta	2044.28	2	1022.14	19.48
Residual and Interaction	577.10	11	52.46	
Total	5865.22	17		

** F Ratio for 99% Significance Level ≥ 7.20

TABLE XV (continued)

2000°F Tensile Elongation (%)

B. 6.3% Al - 1.0 Ti Base

Tungsten	Molybdenum 1.0		4.5		8.0	
1.0	42.1 32.5	(1.0)*	6.0 12.0	(4.5)	8.1 10.8	(8.0)
5.5	17.5 13.2	(4.5)	10.6 8.6	(8.0)	15.6 11.3	(1.0)
10.0	5.5 7.2	(8.0)	10.7 16.2	(1.0)	16.4 15.1	(4.5)

* Numbers in Parentheses are Tantalum Contents

Average Test Values

	1%	4.5% or 5.5%	8.0% or 10.0%
Mo	19.7	10.7	12.9
W	18.6	12.8	11.9
Ta	21.4	13.4	8.5

	Sum of Squares	Degrees of Freedom	Mean Square	F Ratio**
Mo	263.11	2	131.56	2.82
W	159.38	2	79.69	1.71
Ta	511.64	2	255.82	5.49
Residual and Interaction	512.39	11	46.58	
Total	1446.52	17		

** F Ratio for 99% Significance Level $\geq$ 7.20

TABLE XV (continued)

2000°F Tensile Elongation (%)

C. 6.3% Al - 1.8 Ti Base

Tungsten	Molybdenum 1.0		4.5		8.0	
1.0	35.6 42.9	(1.0)*	15.4 22.8	(4.5)	8.6 10.7	(8.0)
5.5	5.2 3.7	(4.5)	5.7 8.3	(8.0)	13.0 12.3	(1.0)
10.0	10.7 6.9	(8.0)	13.2 11.4	(1.0)	17.3 18.5	(4.5)

*Numbers in Parentheses are Tantalum Contents

Average Test Values

	1%	4.5% or 5.5%	8.0% or 10.0%
Mo	17.5	12.8	13.4
W	22.7	8.0	13.0
Ta	21.4	13.8	8.5

	Sum of Squares	Degrees of Freedom	Mean Square	F Ratio**
Mo	78.52	2	39.26	0.78
W	664.50	2	332.25	6.58
Ta	505.58	2	252.79	5.01
Residual and Interaction	555.48	11	50.50	
Total	1804.08	17		

** F Ratio for 99% Significance Level ≥ 7.20

TABLE XVI

1857°F Tensile Elongation (%)

A. 4.5% Al - 1.0 Ti Base

Tungsten	Molybdenum 1.0		Molybdenum 4.5		Molybdenum 8.0	
1.0	31.0 22.7	(1.0)*	21.3 23.9	(4.5)	11.7 8.9	(8.0)
5.5	18.7 21.4	(4.5)	8.1 5.8	(8.0)	14.8 13.3	(1.0)
10.0	9.9 8.5	(8.0)	13.5 11.9	(1.0)	15.0 17.5	(4.5)

* Numbers in Parentheses are Tantalum Contents

Average Test Values

	1%	4.5% or 5.5%	8.0% or 10.0%
Mo	18.7	14.1	13.5
W	19.9	13.7	12.7
Ta	17.9	19.6	8.8

	Sum of Squares	Degrees of Freedom	Mean Square	F Ratio**
Mo	96.62	2	48.31	7.48
W	183.25	2	91.63	14.18
Ta	404.05	2	202.03	31.27
Residual and Interaction	71.10	11	6.46	
Total	755.02	17		

** F Ratio for 99% Significance Level $\geq$ 7.20

TABLE XVI (continued)

1875°F Tensile Elongation (%)

B. 6.3% Al - 1.0 Ti Base

Tungsten	Molybdenum 1.0		4.5		8.0	
1.0	- 22.3	(1.0)*	7.1 7.4	(4.5)	7.8 5.9	(8.0)
5.5	12.8 12.4	(4.5)	7.1 7.3	(8.0)	15.4 11.9	(1.0)
10.0	6.5 5.1	(8.0)	7.2 9.7	(1.0)	12.4 12.9	(4.5)

* Numbers in Parentheses are Tantalum Contents

Average Test Values

	1%	4.5% or 5.5%	8.0% or 10.0%
Mo	13.6	7.6	11.1
W	12.1	11.2	9.0
Ta	14.8	10.8	6.6

	Sum of Squares	Degrees of Freedom	Mean Square	F Ratio**
Mo	106.43	2	53.22	5.29
W	31.53	2	15.77	1.57
Ta	200.96	2	100.48	9.98
Residual and Interaction	110.78	11	10.07	
Total	449.70	17		

**F Ratio for 99% Significance Level ≥ 7.20

TABLE XVI (continued)

1875°F Tensile Elongation (%)

C. 6.3% Al - 1.8 Ti Base

Tungsten	Molybdenum 1.0		4.5		8.0	
1.0	19.5 25.0	(1.0)*	16.1 16.5	(4.5)	6.8 7.3	(8.0)
5.5	3.8 4.5	(4.5)	7.9 7.7	(8.0)	11.7 10.2	(1.0)
10.0	5.1 4.0	(8.0)	9.4 8.9	(1.0)	10.3 9.2	(4.5)

* Numbers in Parentheses are Tantalum Contents

Average Test Values

	1%	4.5% or 5.5%	8.0% or 10.0%
Mo	10.3	11.1	9.3
W	15.2	7.6	7.8
Ta	14.1	10.1	6.5

	Sum of Squares	Degrees of Freedom	Mean Square	F Ratio**
Mo	10.17	2	5.09	0.38
W	223.60	2	111.80	8.45
Ta	175.77	2	87.89	6.64
Residual and Interaction	145.48	11	13.23	
Total	555.02	17		

** F Ratio for 99% Significance Level $\geq$ 7.20

TABLE XVII (continued)

1400°F Tensile Results For Series I Alloys

Alloy	Compositional Aims (Wt.%) Mo	W	Ta	Al	Ti	Ultimate, 1000 psi	0.2% Offset Yield, 1000psi	Elong. (%)	R.A. (%)
X	4.5	1.0	4.5	4.5	1.0	111.0 108.8	83.7 86.4	5.5/5.2 3.5	8.1 7.5
XI	4.5	1.0	4.5	6.3	1.0	132.9 130.0	110.3 110.8	5.5/3.9 3.3	4.7 5.1
XII	4.5	1.0	4.5	6.3	1.8	120.0 130.5	95.1 97.5	5.0/3.0 5.7	6.2 5.5
XIII	4.5	5.5	8.0	4.5	1.0	142.4 136.4	116.6 115.9	3.0/2.7 1.9	3.5 4.9
XIV	4.5	5.5	8.0	6.3	1.0	135.7 129.5	None None	1.0/0.4 0.6	0.8 0.0
XV	4.5	5.5	8.0	6.3	1.8	128.7 136.1	None None	0.0/0.2 0.2	0.0 0.0
XVI	4.5	10.0	1.0	4.5	1.0	116.1 115.8	107.8 107.1	1.0/2.2 1.8	1.6 4.3
XVII	4.5	10.0	1.0	6.3	1.0	127.1 119.6	None None	1.0/0.4 0.4	1.2 1.6
XVIII	4.5	10.0	1.0	6.3	1.8	146.2 144.2	139.0 138.9	2.0/0.3 0.8	1.2 0.8
XIX	8.0	1.0	8.0	4.5	1.0	145.0 145.4	118.8 120.2	4.0/2.9 1.8	3.9 3.9
XX	8.0	1.0	8.0	6.3	1.0	141.9 146.4	None 142.0	1.0/0.8 1.1	0.0 0.8

TABLE XVII (continued)

1400°F Tensile Results For Series I Alloys

	Compositional Aims (Wt.%)								
Alloy	Mo	W	Ta	Al	Ti	Ultimate, 1000 psi	0.2% Offset Yield, 1000psi	Elong. (%)	R.A. (%)
XXI	8.0	1.0	8.0	6.3	1.8	144.9	None	1.0/0.9	0.0
						140.2	None	0.3	0.0
XXII	8.0	5.5	1.0	4.5	1.0	94.5	None	2.5/2.1	0.4
						92.8	None	3.3	0.8
XXIII	8.0	5.5	1.0	6.3	1.0	130.8	129.3	2.0/1.2	0.8
						122.4	None	1.4	1.2
XXIV	8.0	5.5	1.0	6.3	1.8	141.2	139.6	1.5/1.3	0.4
						129.5	None	0.9	0.8
XXV	8.0	10.0	4.5	4.5	1.0	137.1	None	1.5/0.9	0.4
						138.1	None	1.2	0.4
XXVI	8.0	10.0	4.5	6.3	1.0	93.5	None	0.0/0.7	0.0
						93.0	None	0.6	0.0
XXVII	8.0	10.0	4.5	6.3	1.8	98.0	None	0.5/0.6	0.0
						78.2	None	0.5	0.0

TABLE XVIII

Room Temperature Tensile Results For Series I Alloys

Alloy	Compositional Aims (Wt.%) Mo	W	Ta	Al	Ti	Ultimate, 1000 psi	0.2% Offset Yield, 1000psi	Elong. (%)	R.A. (%)
I	1.0	1.0	1.0	4.5	1.0	104.7 105.2	85.4 84.5	5.0/4.1* 5.6	5.9 10.2
II	1.0	1.0	1.0	6.3	1.0	106.3 95.9	88.6 81.4	9.0/8.0 10.4	15.6 18.8
III	1.0	1.0	1.0	6.3	1.8	123.1 117.2	94.2 93.3	9.0/9.3 8.0	12.7 12.3
IV	1.0	5.5	4.5	4.5	1.0	128.1 141.8	102.8 113.0	6.0/5.6 8.0	8.5 13.1
V	1.0	5.5	4.5	6.3	1.0	135.1 133.6	133.4 114.0	7.0/7.1 6.8	7.1 7.0
VI	1.0	5.5	4.5	6.3	1.8	137.8 144.4	120.7 111.0	5.5/4.6 10.1	6.3 9.2
VII	1.0	10.0	8.0	4.5	1.0	148.2 148.2	130.1 None	5.0/3.9 5.0	7.4 8.9
VIII	1.0	10.0	8.0	6.3	1.0	149.4 144.4	143.1 143.2	1.0/0.2 0.3	1.6 1.2
IX	1.0	10.0	8.0	6.3	1.8	135.4 145.2	None None	1.0/0.3 0.2	1.2 0.8
X	4.5	1.0	4.5	4.5	1.0	120.6 125.3	97.9 99.5	6.5/6.2 6.9	10.0 9.3

*First value is elongation in 1" gage length. Second value is radius to radius elongation

TABLE XVIII (continued)

Room Temperature Tensile Results For Series I Alloys

Alloy	Compositional Aims (Wt.%) Mo	W	Ta	Al	Ti	Ultimate, 1000 psi	0.2% Offset Yield, 1000psi	Elong. (%)	R.A. (%)
XI	4.5	1.0	4.5	6.3	1.0	124.9 123.1	109.5 106.7	6.0/5.1 6.5	6.9 10.8
XII	4.5	1.0	4.5	6.3	1.8	134.1 137.1	109.9 109.1	7.0/7.7 9.3	9.7 11.0
XIII	4.5	5.5	8.0	4.5	1.0	143.9 142.7	127.9 126.5	3.0/3.4 3.7	7.4 5.9
XIV	4.5	5.5	8.0	6.3	1.0	150.5 153.2	None 151.8	1.0/0.6 0.4	1.6 1.6
XV	4.5	5.5	8.0	6.3	1.8	137.6 138.6	None None	0.0/0.0 0.3	0.0 0.4
XVI	4.5	10.0	1.0	4.5	1.0	125.9 124.5	119.9 119.0	3.0/1.9 2.2	4.2 2.7
XVII	4.5	10.0	1.0	6.3	1.0	132.5 132.4	129.7 129.0	2.0/2.1 1.8	2.0 1.6
XVIII	4.5	10.0	1.0	6.3	1.8	146.7 150.2	146.3 147.8	1.0/0.7 1.4	1.2 2.4
XIX	8.0	1.0	8.0	4.5	1.0	142.1 143.9	127.3 126.3	3.0/2.5 3.9	5.8 7.8
XX	8.0	1.0	8.0	6.3	1.0	151.0 149.0	None 147.8	1.0/1.2 1.0	1.6 1.2
XXI	8.0	1.0	8.0	6.3	1.8	148.6 147.3	None None	0.0/0.0 0.6	0.0 0.0

TABLE XVIII (continued)

Room Temperature Tensile Results For Series I Alloys

	Compositional Aims (Wt.%)								
Alloy	Mo	W	Ta	Al	Ti	Ultimate, 1000 psi	0.2% Offset Yield, 1000psi	Elong. (%)	R.A. (%)
XXII	8.0	5.5	1.0	4.5	1.0	116.7	116.1	1.0/1.4	1.6
						116.7	None	1.8	6.3
XXIII	8.0	5.5	1.0	6.3	1.0	129.7	125.5	1.0/1.7	3.6
						133.3	128.2	1.9	2.4
XXIV	8.0	5.5	1.0	6.3	1.8	156.2	None	0.0/0.5	0.0
						152.4	None	0.8	0.0
XXV	8.0	10.0	4.5	4.5	1.0	152.4	150.8	0.0/0.3	0.0
						152.2	151.8	0.8	0.8
XXVI	8.0	10.0	4.5	6.3	1.0	101.2	None	0.0/0.0	0.0
						102.9	None	0.0	0.0
XXVII	8.0	10.0	4.5	6.3	1.8	89.7	None	3.0/1.6	0.9
						94.0	None	0.6	0.4

TABLE XIX

1400°F Tensile Strength, 1000 psi

A. 4.5% Al - 1.0 Ti Base

Tungsten	Molybdenum 1.0		Molybdenum 4.5		Molybdenum 8.0	
1.0	92.0 88.1	(1.0)*	111.0 108.8	(4.5)	145.0 145.4	(8.0)
5.5	106.0 98.7	(4.5)	142.4 136.4	(8.0)	94.5 92.8	(1.0)
10.0	147.4 148.2	(8.0)	116.1 115.8	(1.0)	137.1 138.1	(4.5)

*Numbers in Parentheses are Tantalum Contents

Average Test Values

	1%	4.5% or 5.5%	8.0% or 10.0%
Mo	113.4	121.8	125.5
W	115.1	111.8	133.8
Ta	99.9	116.6	144.1

	Sum of Squares	Degrees of Freedom	Mean Square	F Ratio**
Mo	459.34	2	229.67	17.14
W	1689.54	2	844.77	63.04
Ta	5990.47	2	2995.24	223.53
Residual and Interaction	147.43	11	13.40	
Total	8286.78	17		

** F Ratio for 99% Significance Level ≥ 7.20

TABLE XIX (continued)

1400°F Tensile Strength, 1000 psi

B. 6.3% Al - 1.0 Ti Base

Tungsten	Molybdenum 1.0		4.5		8.0	
1.0	84.5 99.8	(1.0)*	132.9 130.0	(4.5)	141.9 146.4	(8.0)
5.5	139.7 136.8	(4.5)	135.7 129.5	(8.0)	130.7 122.4	(1.0)
10.0	145.6 139.2	(8.0)	127.1 119.6	(1.0)	93.5 93.0	(4.5)

*Numbers in Parentheses are Tantalum Contents

Average Test Values

	1%	4.5% or 5.5%	8.0% or 10.0%
Mo	124.3	129.1	121.3
W	122.6	132.5	119.7
Ta	114.0	121.0	139.7

	Sum of Squares	Degrees of Freedom	Mean Square	F Ratio**
Mo	186.98	2	93.49	0.30
W	540.06	2	270.03	0.86
Ta	2119.92	2	1059.96	3.37
Residual and Interaction	3464.29	11	314.94	
Total	6311.25	17		

** F Ratio for 99% Significance Level $\geq$ 7.20

TABLE XIX (continued)

1400°F Tensile Strength, 1000 psi

C. 6.3% Al - 1.8 Ti Base

Tungsten	Molybdenum 1.0		4.5		8.0	
1.0	116.3 114.0	(1.0)*	120.0 130.5	(4.5)	144.9 140.2	(8.0)
5.5	142.4 140.4	(4.5)	128.7 136.1	(8.0)	141.2 129.5	(1.0)
10.0	135.3 143.2	(8.0)	146.2 144.2	(1.0)	98.0 78.2	(4.5)

*Numbers in Parentheses are Tantalum Contents

Average Test Values

	1%	4.5% or 5.5%	8.0% or 10.0%
Mo	131.9	134.3	122.0
W	127.7	136.4	124.2
Ta	131.9	118.3	138.1

	Sum of Squares	Degrees of Freedom	Mean Square	F Ratio**
Mo	510.15	2	255.08	0.81
W	473.26	2	236.63	0.76
Ta	1234.11	2	617.06	1.97
Residual and Interaction	3446.27	11	313.30	
Total	5663.79	17		

** F Ratio for 99% Significance Level $\geq$ 7.20

TABLE XX

Room Temperature Tensile Strength, 1000 psi

A. 4.5% Al - 1.0 Ti Base

Tungsten	Molybdenum 1.0		4.5		8.0	
1.0	104.7 105.2	(1.0)*	120.6 125.3	(4.5)	142.1 143.9	(8.0)
5.5	128.1 141.8	(4.5)	143.9 142.7	(8.0)	116.7 116.7	(1.0)
10.0	148.2 148.2	(8.0)	125.9 124.5	(1.0)	152.4 152.2	(4.5)

* Number in Parentheses are Tantalum Contents

Average Test Values

	1%	4.5% or 5.5%	8.0% or 10.0%
Mo	129.4	130.5	137.3
W	123.6	131.7	141.9
Ta	115.6	136.7	144.8

	Sum of Squares	Degrees of Freedom	Mean Square	F Ratio**
Mo	223.28	2	111.64	6.21
W	1006.01	2	503.01	27.99
Ta	2730.28	2	1365.14	75.97
Residual and Interaction	197.70	11	17.97	
Total	4157.27	17		

** F Ratio for 99% Significance Level $\geq$ 7.20

TABLE XX (continued)

Room Temperature Tensile Strength, 1000 psi

B. 6.3% Al - 1.0 Ti Base

Tungsten	Molybdenum 1.0		4.5		8.0	
1.0	106.3 95.9	(1.0)*	124.9 123.1	(4.5)	151.0 149.0	(8.0)
5.5	135.1 133.6	(4.5)	150.5 153.2	(8.0)	129.7 133.3	(1.0)
10.0	149.4 144.4	(8.0)	132.5 132.4	(1.0)	101.2 102.9	(4.5)

* Number in Parentheses are Tantalum Contents

Average Test Values

	1%	4.5% or 5.5%	8.0% or 10.0%
Mo	127.5	136.1	127.9
W	125.0	139.2	127.1
Ta	121.7	120.1	149.6

	Sum of Squares	Degrees of Freedom	Mean Square	F Ratio**
Mo	286.10	2	143.05	1.07
W	704.93	2	352.47	2.63
Ta	3296.23	2	1648.12	12.27
Residual and Interaction	1476.96	11	134.27	
Total	5764.22	17		

** F Ratio for 99% Significance Level ≥ 7.20

TABLE XX (continued)

Room Temperature Tensile Strength, 1000 psi

C. 6.3% Al - 1.8 Ti Base

Tungsten	Molybdenum 1.0		4.5		8.0	
1.0	123.1 117.2	(1.0)*	134.1 137.1	(4.5)	148.6 147.3	(8.0)
5.5	137.8 144.4	(4.5)	137.6 138.6	(8.0)	156.2 152.4	(1.0)
10.0	135.4 145.2	(8.0)	146.7 150.2	(1.0)	89.7 94.0	(4.5)

* Numbers in Parentheses are Tantalum Contents

Average Test Values

	1%	4.5% or 5.5%	8.0% or 10.0%
Mo	133.9	140.7	131.4
W	134.6	144.5	126.9
Ta	141.0	122.9	142.1

	Sum of Squares	Degrees of Freedom	Mean Square	F Ratio**
Mo	281.49	2	140.75	0.48
W	937.80	2	468.90	1.59
Ta	1401.49	2	700.75	2.37
Residual and Interaction	3250.74	11	295.52	
Total	5871.52	17		

** F Ratio for 99% Significance Level $\geq$ 7.20

TABLE XXI

1400°F Tensile Elongation (%)

A. 4.5% Al - 1.0 Ti Base

Tungsten	Molybdenum 1.0		Molybdenum 4.5		Molybdenum 8.0	
1.0	2.9 1.8	(1.0)*	5.2 3.5	(4.5)	2.9 1.8	(8.0)
5.5	2.8 2.8	(4.5)	2.7 1.9	(8.0)	2.1 3.3	(1.0)
10.0	4.8 2.8	(8.0)	2.2 1.8	(1.0)	0.9 1.2	(4.5)

* Numbers in Parentheses are Tantalum Contents.

Average Test Values

	1%	4.5% or 5.5%	8.0% or 10.0%
Mo	2.98	2.88	2.03
W	3.02	2.60	2.28
Ta	2.35	2.73	2.82

	Sum of Squares	Degrees of Freedom	Mean Square	F Ratio**
Mo	3.27	2	1.64	1.19
W	1.62	2	0.81	0.59
Ta	0.75	2	0.38	0.28
Residual and Interaction	15.22	11	1.38	
Total	20.86	17		

** F Ratio for 99% Significance Level ≥ 7.20

TABLE XXI (continued)

1400°F Tensile Elongation (%)

B. 6.3% Al - 1.0 Ti Base

Tungsten	Molybdenum 1.0		4.5		8.0	
1.0	3.0 2.0	(1.0)*	3.9 3.3	(4.5)	0.8 1.1	(8.0)
5.5	4.2 3.1	(4.5)	0.4 0.6	(8.0)	1.2 1.4	(1.0)
10.0	0.6 1.0	(8.0)	0.4 0.4	(1.0)	0.7 0.6	(4.5)

* Numbers in Parentheses are Tantalum Contents

Average Test Values

	1%	4.5% or 5.5%	8.0% or 10.0%
Mo	2.32	1.50	0.97
W	2.35	1.82	0.62
Ta	1.40	2.63	0.75

	Sum of Squares	Degrees of Freedom	Mean Square	F Ratio**
Mo	5.54	2	2.77	13.19
W	9.46	2	4.73	22.52
Ta	11.56	2	5.78	27.52
Residual and Interaction	2.33	11	0.21	
Total	28.89	17		

** F Ratio for 99% Significance Level $\geq$7.20

TABLE XXI (continued)

1400°F Tensile Elongation (%)

C. 6.3% Al - 1.8 Ti Base

Tungsten	Molybdenum 1.0		Molybdenum 4.5		Molybdenum 8.0	
1.0	4.6 3.3	(1.0)*	3.0 5.7	(4.5)	0.9 0.3	(8.0)
5.5	2.6 3.0	(4.5)	0.2 0.2	(8.0)	1.3 0.9	(1.0)
10.0	0.3 0.9	(8.0)	0.3 0.8	(1.0)	0.6 0.5	(4.5)

*Numbers in Parentheses are Tantalum Contents

Average Test Values

	1%	4.5% or 5.5%	8.0% or 10.0%
Mo	2.45	1.70	0.75
W	2.97	1.37	0.57
Ta	1.87	2.57	0.47

	Sum of Squares	Degress of Freedom	Mean Square	F Ratio**
Mo	8.72	2	4.36	7.27
W	17.93	2	8.97	14.95
Ta	13.73	2	6.87	11.45
Residual and Interaction	6.62	11	0.60	
Total	47.00	17		

** F Ratio for 99% Significance Level $\geq$ 7.20

TABLE XXII

Room Temperature Tensile Elongation (%)

A. 4.5% Al - 1.0 Ti Base

Tungsten	Molybdenum 1.0		4.5		8.0	
1.0	4.1 5.6	(1.0)*	6.2 6.9	(4.5)	2.5 3.9	(8.0)
5.5	5.6 8.0	(4.5)	3.4 3.7	(8.0)	1.7 1.9	(1.0)
10.0	3.9 5.0	(8.0)	1.9 2.2	(1.0)	0.3 0.8	(4.5)

* Numbers in Parentheses are Tantalum Contents

Average Test Values

	1%	4.5% or 5.5%	8.0% or 10.0%
Mo	5.4	4.1	1.9
W	4.9	4.1	2.4
Ta	2.9	4.6	3.7

	Sum of Squares	Degrees of Freedom	Mean Square	F Ratio**
Mo	37.89	2	18.95	18.05
W	19.79	2	9.90	9.43
Ta	9.02	2	4.51	4.30
Residual and Interaction	11.60	11	1.05	
Total	78.30	17		

** F Ratio for 99% Significance Level ≥ 7.20

TABLE XXII (continued)

Room Temperature Tensile Elongation (%)

B. 6.3% Al - 1.0 Ti Base

Tungsten	Molybdenum 1.0		4.5		8.0	
1.0	8.0 10.4	(1.0)*	5.1 6.5	(4.5)	1.2 1.0	(8.0)
5.5	7.1 6.8	(4.5)	0.6 0.4	(8.0)	1.4 1.8	(1.0)
10.0	0.2 0.3	(8.0)	2.1 1.8	(1.0)	0.0 0.0	(4.5)

*Numbers in Parentheses are Tantalum Contents

Average Test Values

	1%	4.5% or 5.5%	8.0% or 10.0%
Mo	5.5	2.8	0.9
W	5.4	3.0	0.7
Ta	4.3	4.3	0.6

	Sum of Squares	Degrees of Freedom	Mean Square	F Ratio**
Mo	63.32	2	31.66	55.54
W	64.41	2	32.21	56.51
Ta	52.81	2	26.41	46.33
Residual and Interaction	6.24	11	0.57	
Total	186.78	17		

** F Ratio for 99% Significance Level ≥ 7.20

TABLE XXII (continued)

Room Temperature Tensile Elongation (%)

C. 6.3% Al - 1.8 Ti Base

Tungsten	Molybdenum 1.0		Molybdenum 4.5		Molybdenum 8.0	
1.0	9.3 8.0	(1.0)*	7.7 9.3	(4.5)	0.0 0.6	(8.0)
5.5	4.6 10.1	(4.5)	0.0 0.3	(8.0)	0.5 0.8	(1.0)
10.0	0.3 0.2	(8.0)	0.7 1.4	(1.0)	1.6 0.6	(4.5)

* Numbers in Parentheses are Tantalum Contents

Average Test Values

	1%	4.5% or 5.5%	8.0% or 10.0%
Mo	5.4	3.2	0.7
W	5.8	2.7	0.8
Ta	3.5	5.7	0.2

	Sum of Squares	Degrees of Freedom	Mean Square	F Ratio**
Mo	67.35	2	33.68	19.81
W	76.90	2	38.45	22.62
Ta	89.07	2	44.54	26.20
Residual and Interaction	18.74	11	1.70	
Total	252.06	17		

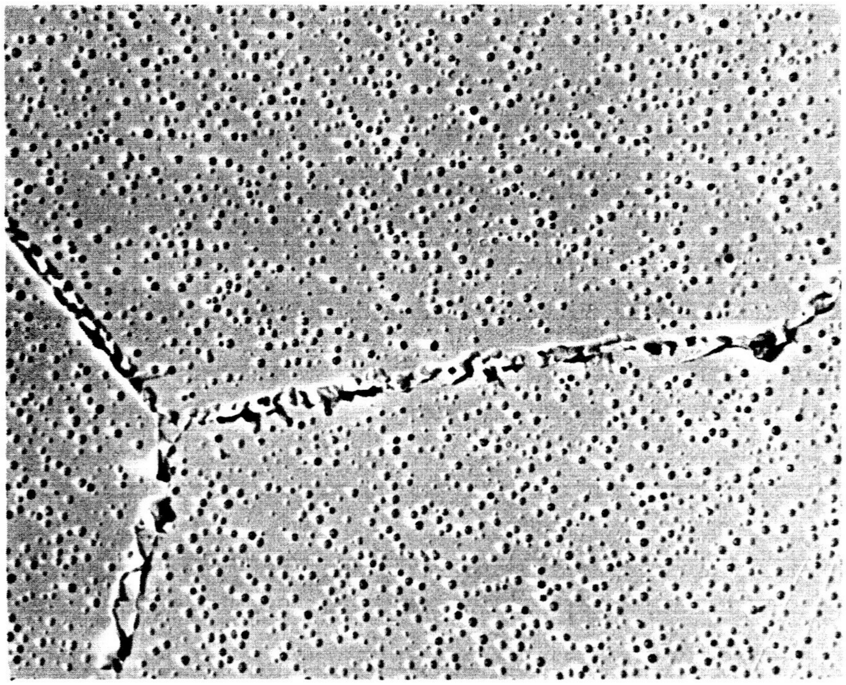
** F Ratio for 99% Significance Level $\geq$ 7.20

TABLE XVII

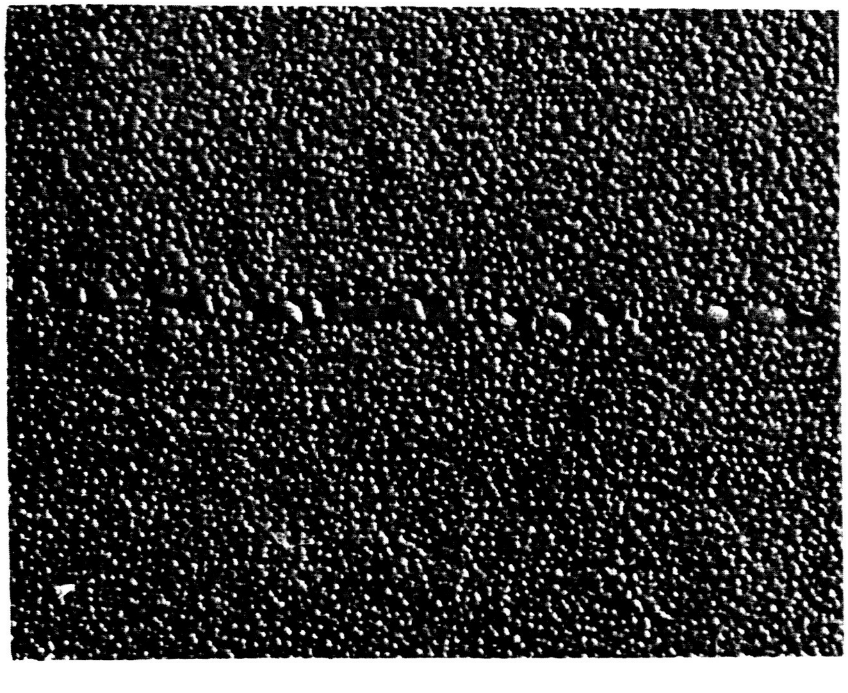
1400°F Tensile Results For Series I Alloys

Alloy	Compositional Aims (Wt.%) Mo	W	Ta	Al	Ti	Ultimate, 1000psi	0.2% Offset Yield, 1000psi	Elong. (%)	R.A. (%)
I	1.0	1.0	1.0	4.5	1.0	92.0 88.1	65.7 72.2	2.0/2.9* 1.8	0.8 2.4
II	1.0	1.0	1.0	6.3	1.0	84.5 99.8	79.7 78.0	2.5/3.0 2.0	5.1 4.3
III	1.0	1.0	1.0	6.3	1.8	116.3 114.0	87.8 88.1	5.0/4.6 3.3	7.0 8.6
IV	1.0	5.5	4.5	4.5	1.0	106.0 98.7	85.3 85.5	4.0/2.8 2.8	7.7 5.5
V	1.0	5.5	4.5	6.3	1.0	139.7 136.8	114.4 89.8	5.0/4.2 3.1	4.7 4.0
VI	1.0	5.5	4.5	6.3	1.8	142.4 140.4	120.0 117.0	4.0/2.6 3.0	3.9 6.2
VII	1.0	10.0	8.0	4.5	1.0	147.4 148.2	115.7 118.2	6.0/4.8 2.8	7.3 4.7
VIII	1.0	10.0	8.0	6.3	1.0	145.6 139.2	None 137.9	1.5/0.6 1.0	0.8 0.8
IX	1.0	10.0	8.0	6.3	1.8	135.3 143.2	None 137.1	0.0/0.3 0.9	0.0 0.0

*First Value is Elongation in 1" Gage Length. Second Value is Radius to Radius Elongation



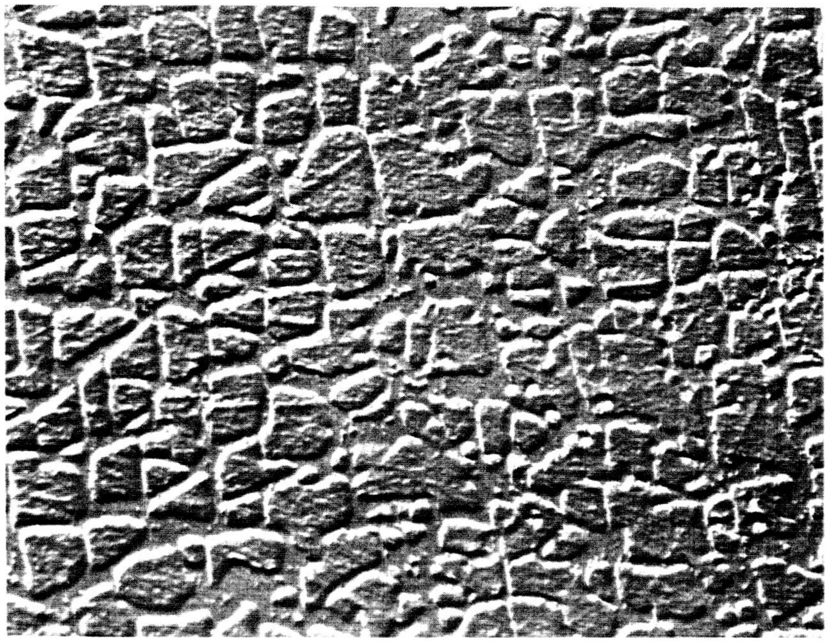
A. Waspaloy - Solution Treated at 2100°F, Aged at 1700°F, 6000X Magnification



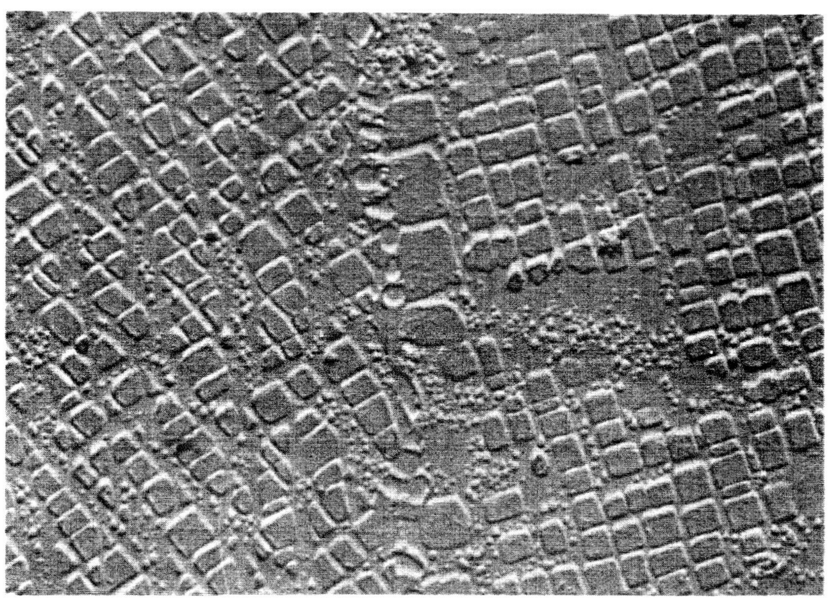
B. Udimet 700 - Solution Treated at 2150°F, Water Quenched, 15,000X Magnification

Figure 1. Examples of Gamma Prime Formations in Nickel-Base Superalloys.

3792



A. Gamma Prime Formations in Cast TRW 1900.



B. Coarse and Fine Gamma Prime Formations in Heat Treated Nimonic 115.

Figure 2. Blocky Intragranular Gamma Prime Formations in Nickel-Base Superalloys. 10,000X Magnification

9873

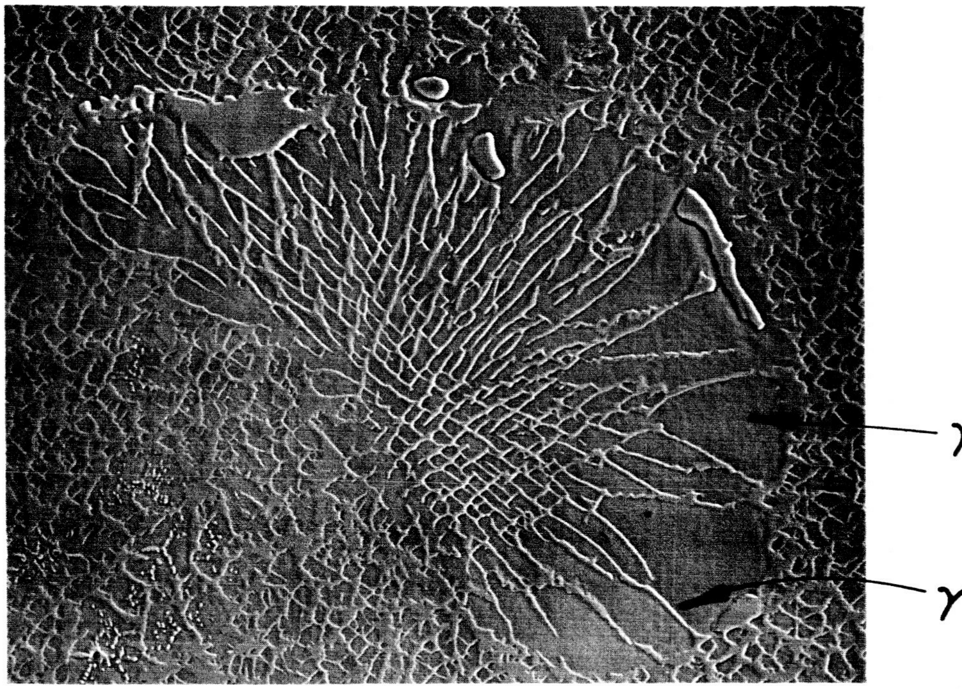
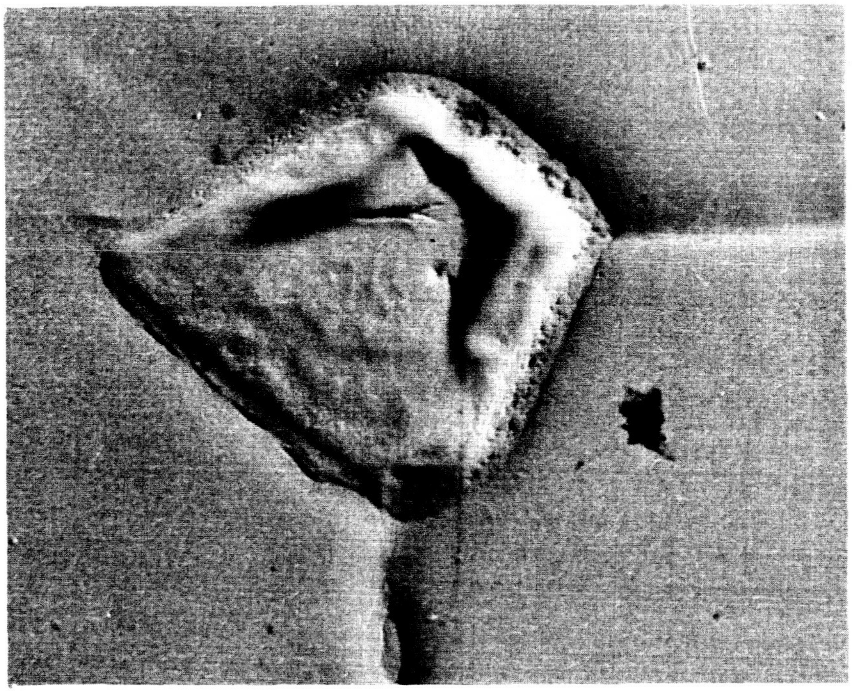
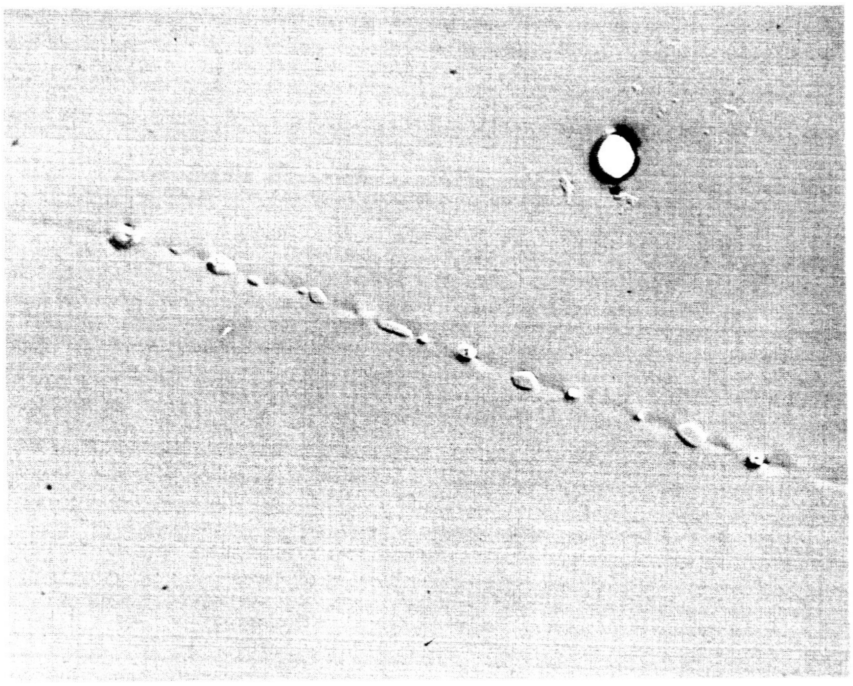


Figure 3. Eutectic Structure of Gamma Prime and Gamma Phases ("Primary Gamma Prime") in Cast Nickel-Base Superalloy TRW 1800.
3400X Magnification



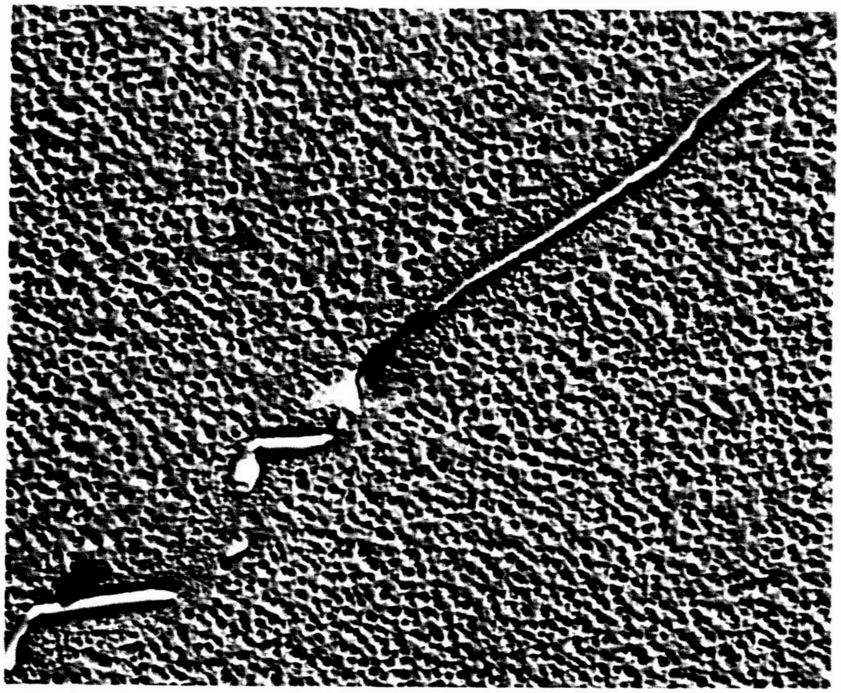
A. Massive MC Carbide Particle. 10,000 Magnification



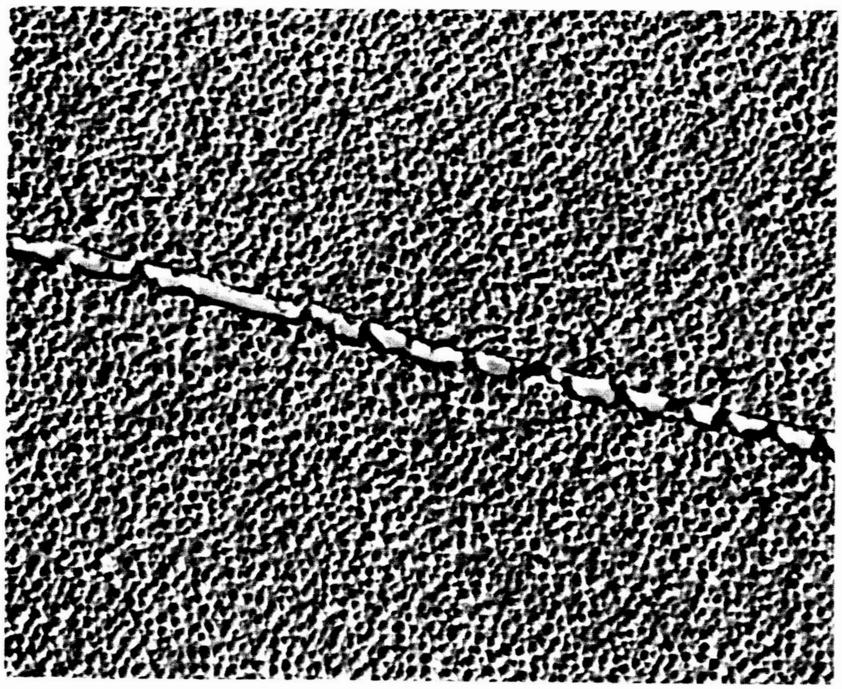
B. Grain Boundary MC Particles. 6000X Magnification

Figure 4. MC Carbide Formations in Waspaloy.

3794



Platelet M_{23} C_6 Formation



Discrete Particle M_{23} C_6 Formation

Figure 5. Grain Boundary $M_{23}C_6$ Formation in Inconel 700.
Both 15,000X Magnification

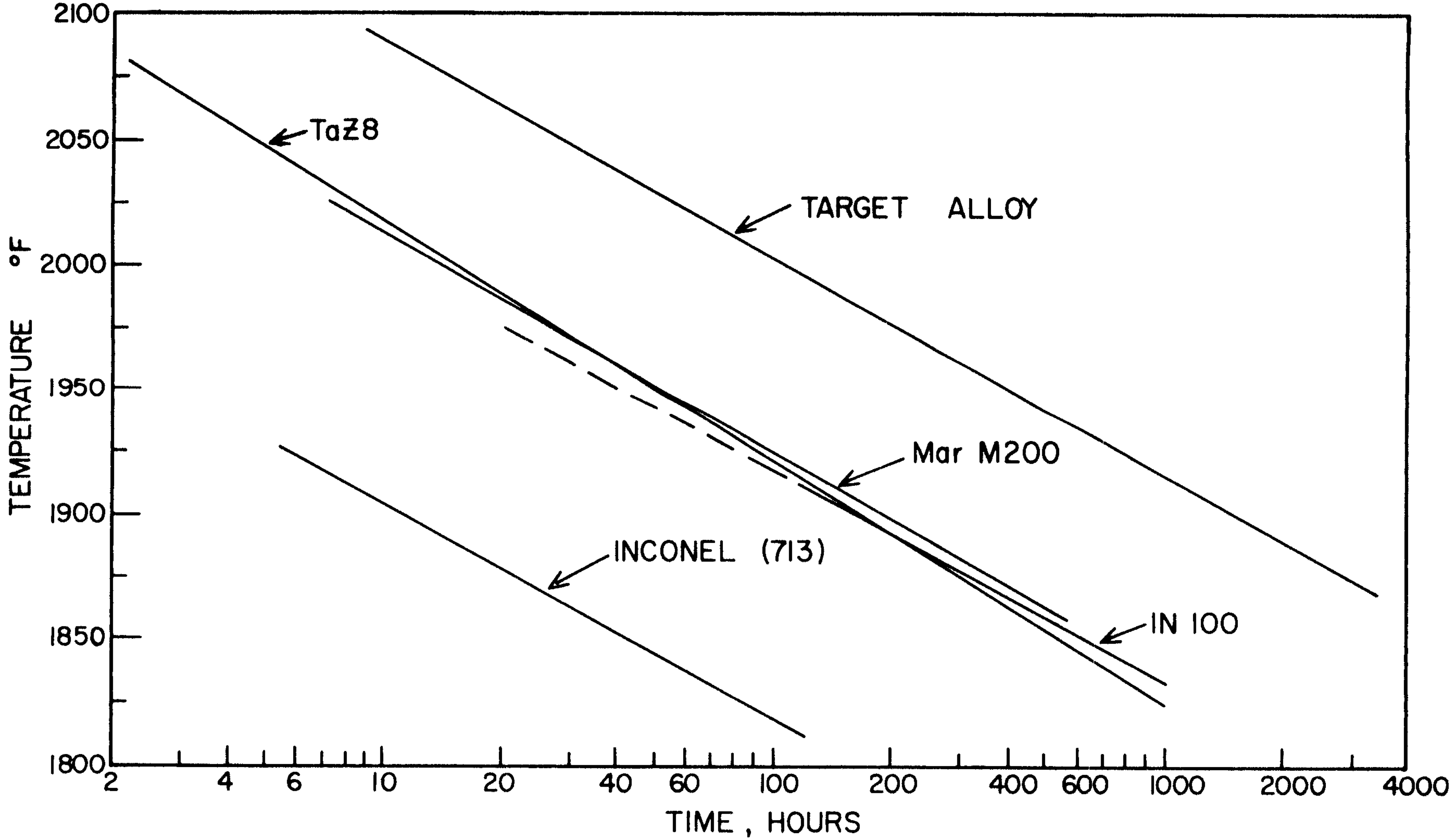


Figure 6. Stress Rupture Properties at 15,000 psi of Existent Cast Nickel-Base Superalloys in Relation to Target Alloy. (After Freche and Waters)[15]

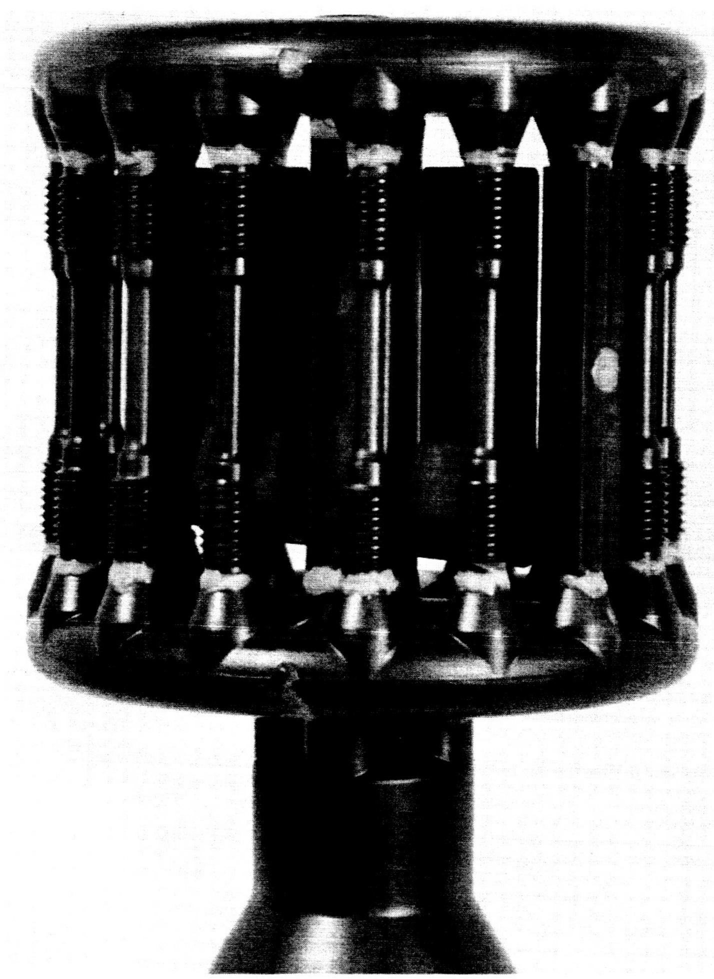


Figure 7. Threaded Tensile Bar Cluster Similar to Tapered Tensile Bar Design Used to Cast NASA Nickel-Base Superalloys.

744

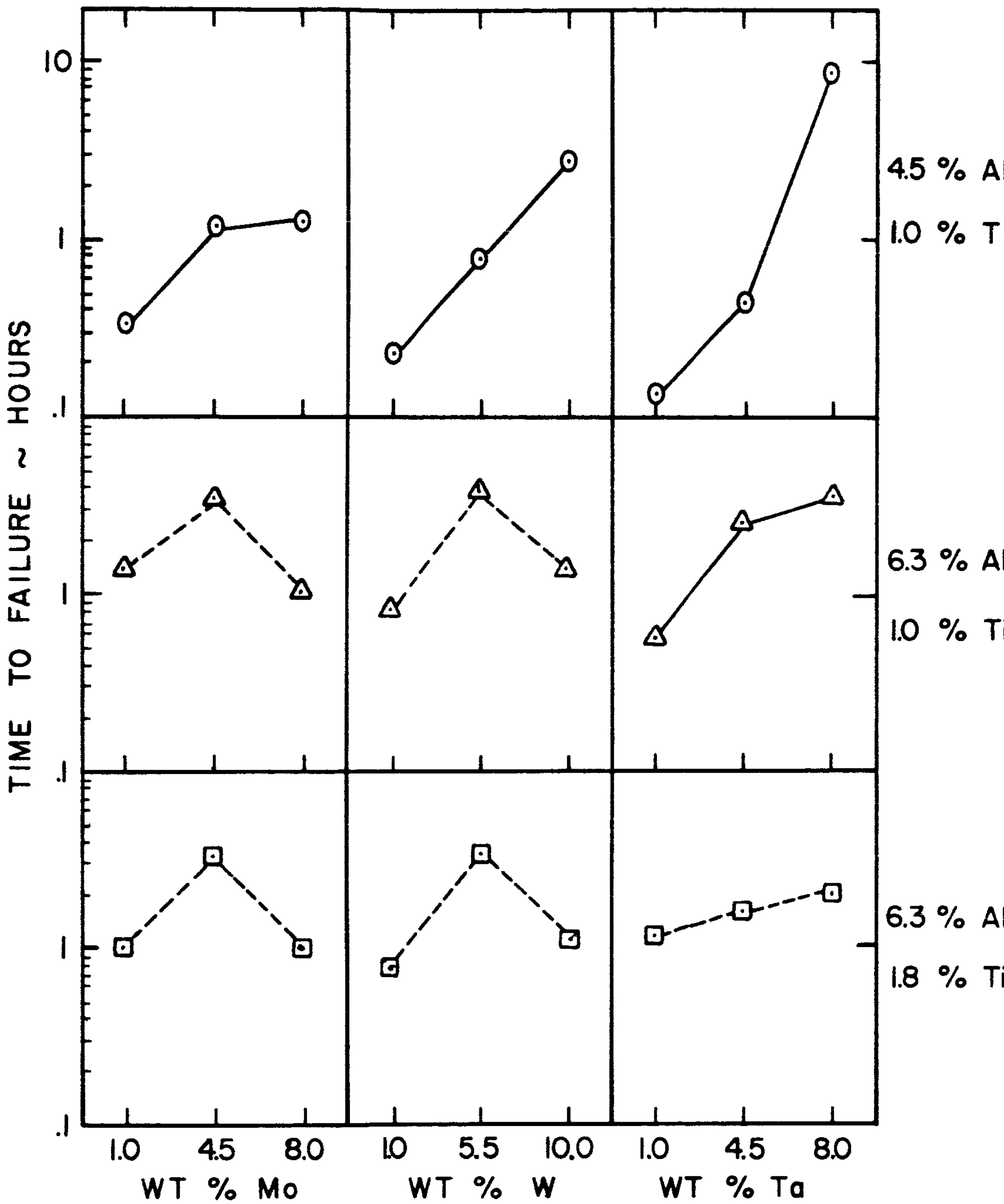


Figure 8. Average 2000°F, 15,000 psi Stress Rupture Life for Nickel-Base Superalloys Containing Various Molybdenum, Tungsten, and Tantalum Levels and Having Aluminum and Titanium Contents Shown. (Solid Lines are Statistically Significant.)

9862

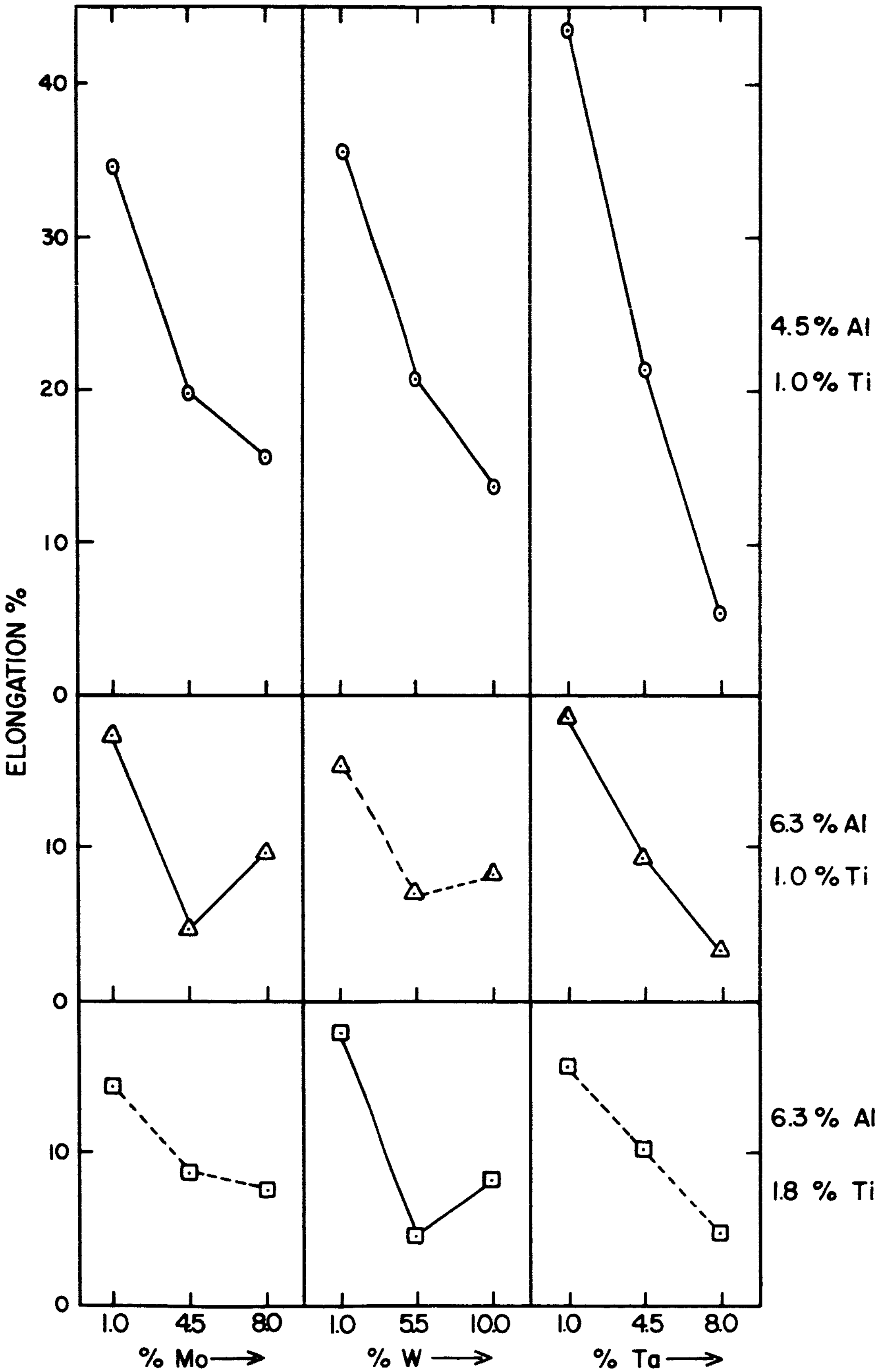


Figure 9. Average 2000°F, 15,000 psi Stress Rupture Elongation for Nickel-Base Superalloys Containing Various Molybdenum, Tungsten, and Tantalum Levels and Having Aluminum and Titanium Contents Shown. (Solid Lines are Statistically Significant.)

9876

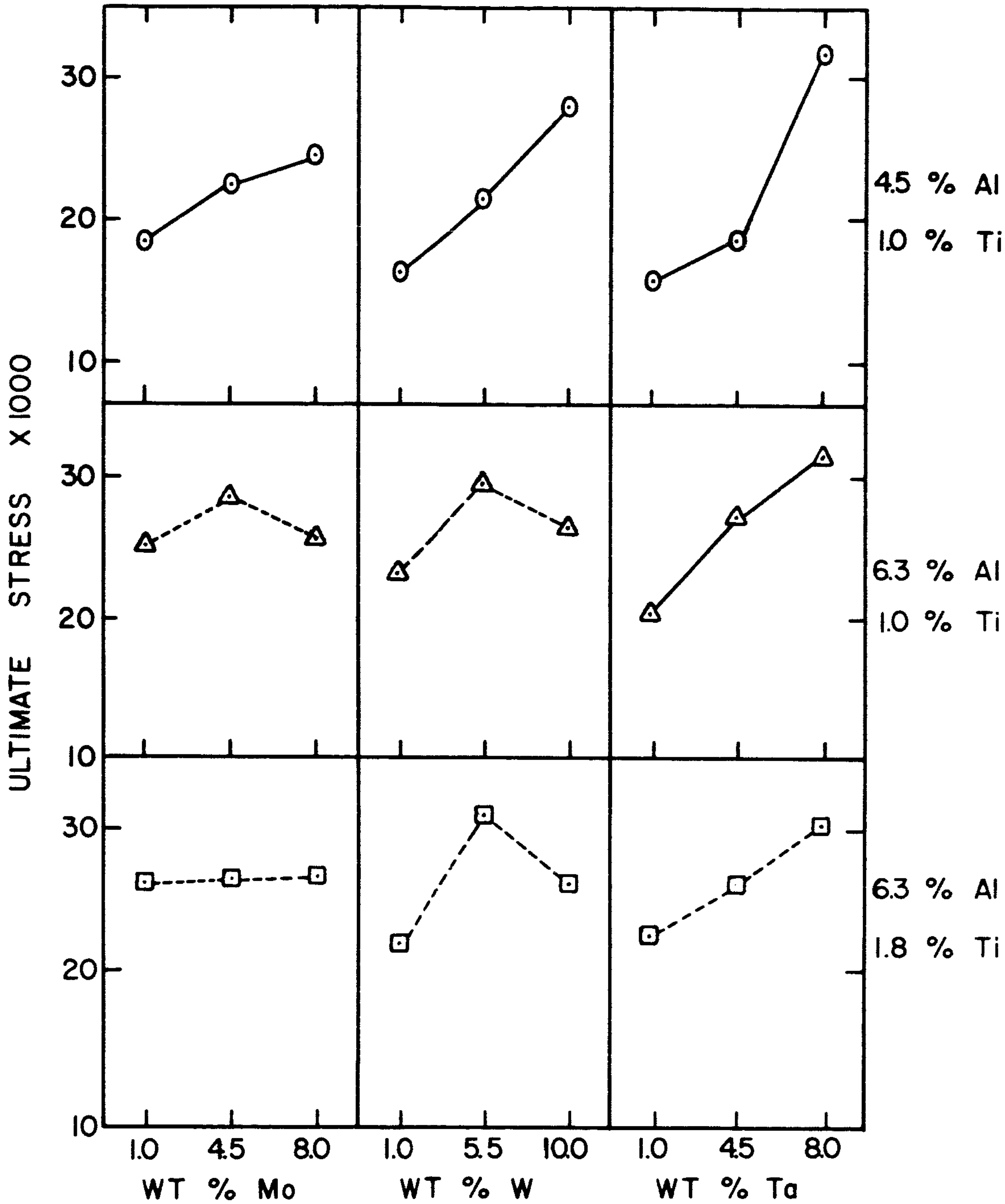


Figure 10. Average 2000°F Tensile Strength for Nickel-Base Superalloys Containing Various Molybdenum, Tungsten, and Tantalum Levels and Having Aluminum and Titanium Contents Shown. (Solid Lines are Statistically Significant.)

9859

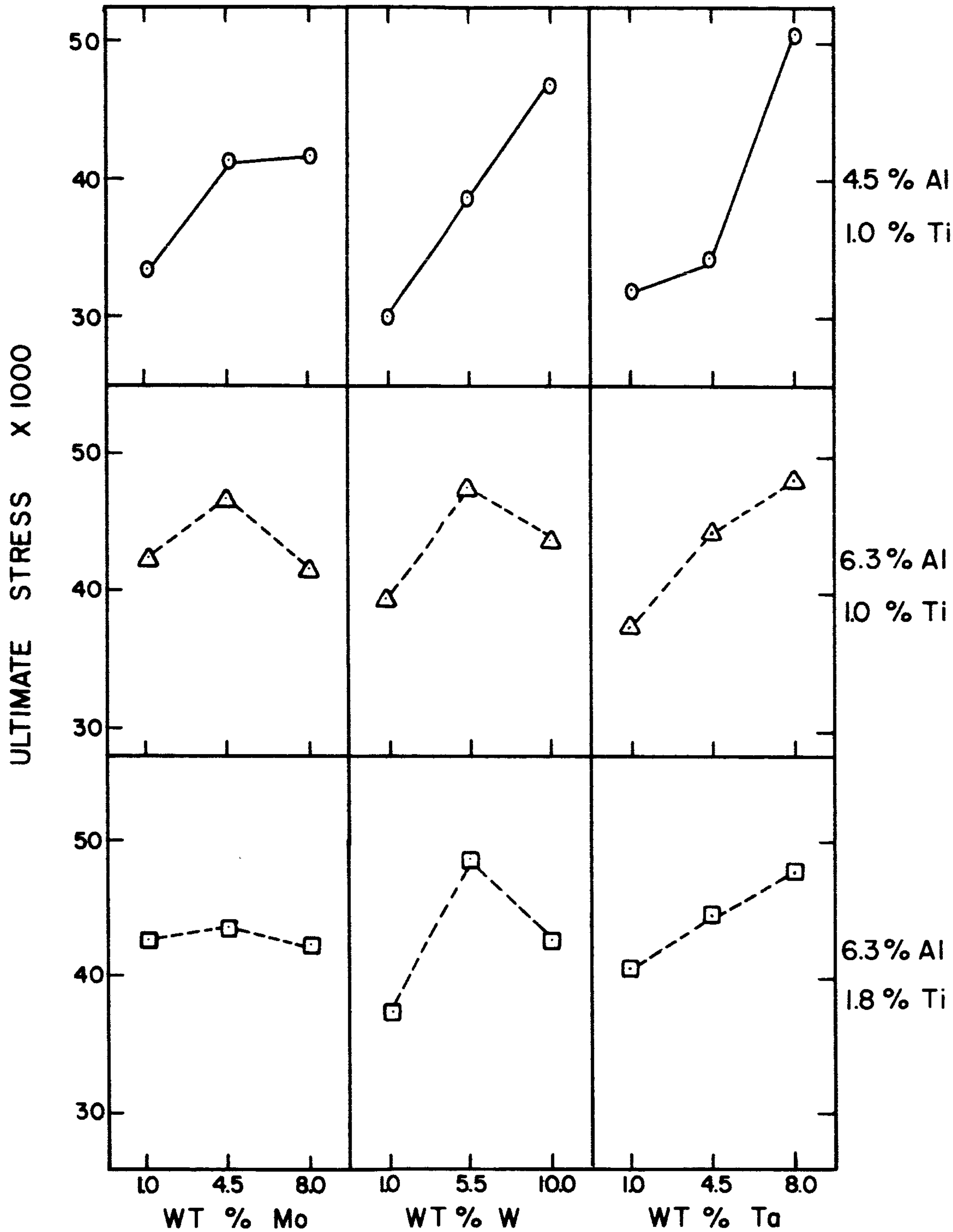


Figure 11. Average 1875°F Tensile Strength for Nickel-Base Superalloys Containing Various Molybdenum, Tungsten, and Tantalum Levels and Having Aluminum and Titanium Contents Shown. (Solid Lines are Statistically Significant.)

9861

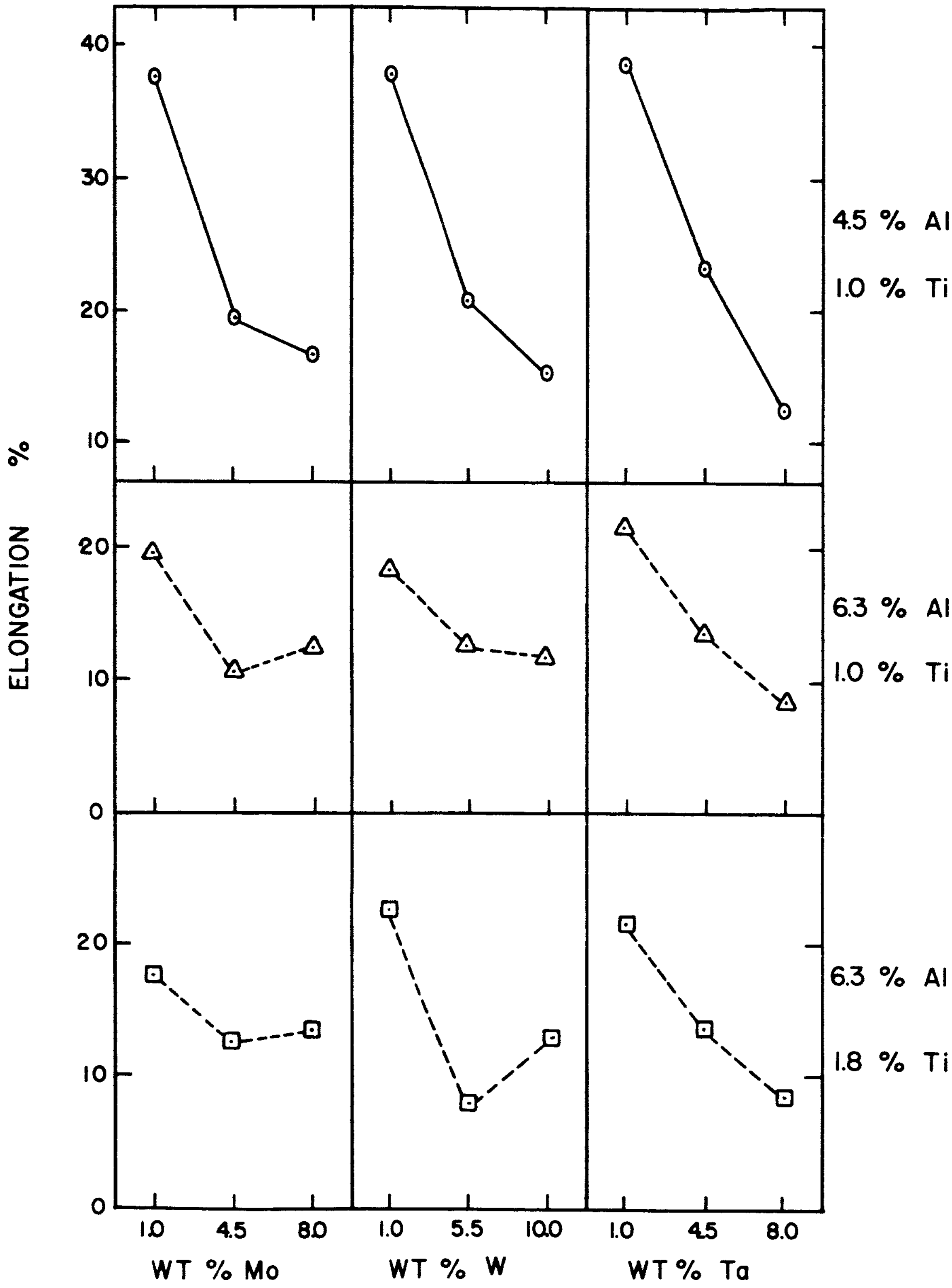


Figure 12. Average 2000°F Tensile Elongation for Nickel-Base Superalloys Containing Various Molybdenum, Tungsten, and Tantalum Levels and Having Aluminum and Titanium Contents Shown. (Solid Lines are Statistically Significant.)

9865

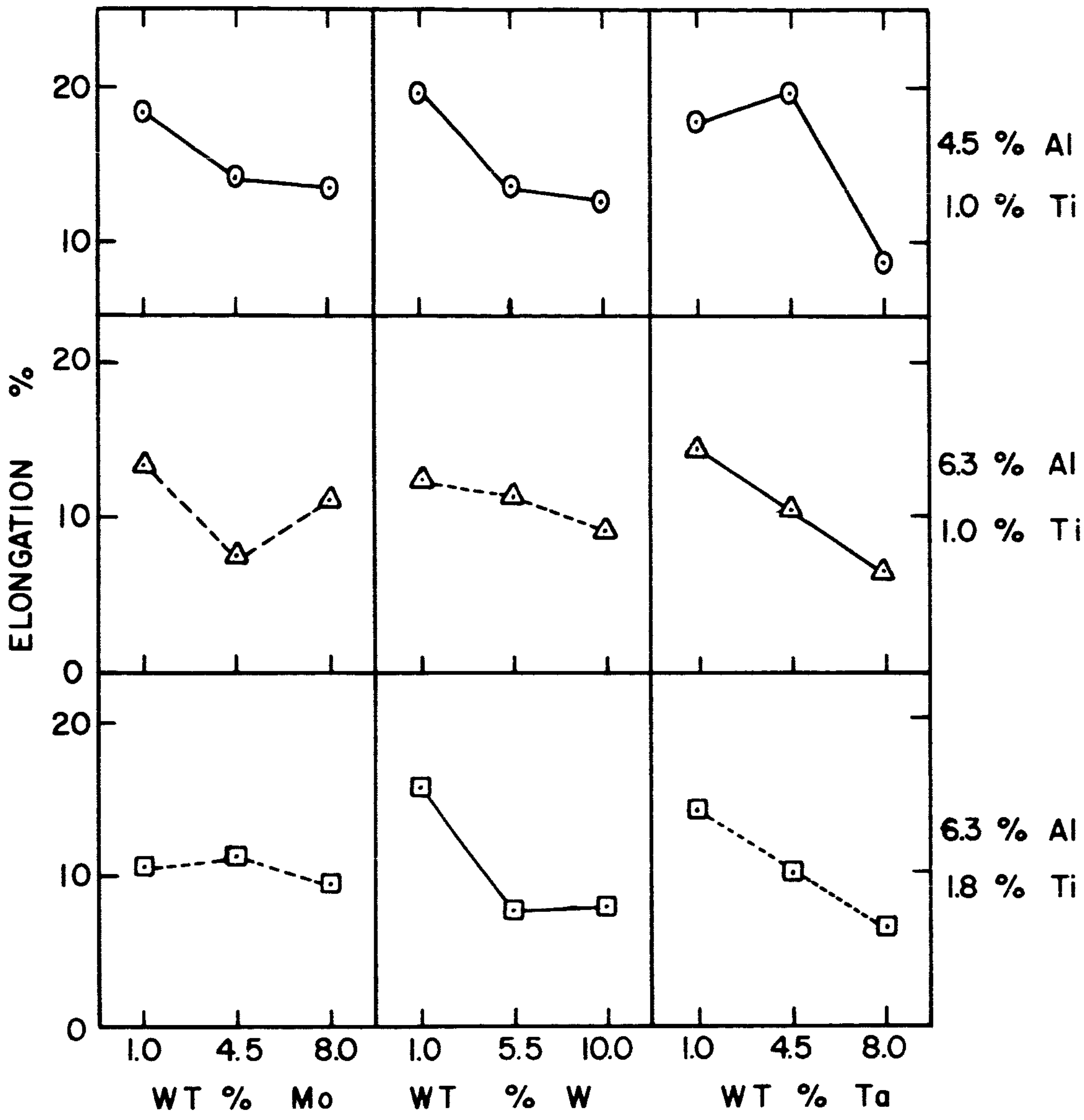


Figure 13. Average 1875°F Tensile Elongation for Nickel-Base Superalloys Containing Various Molybdenum, Tungsten, and Tantalum Levels and Having Aluminum and Titanium Contents Shown. (Solid Lines are Statistically Significant.)

9867

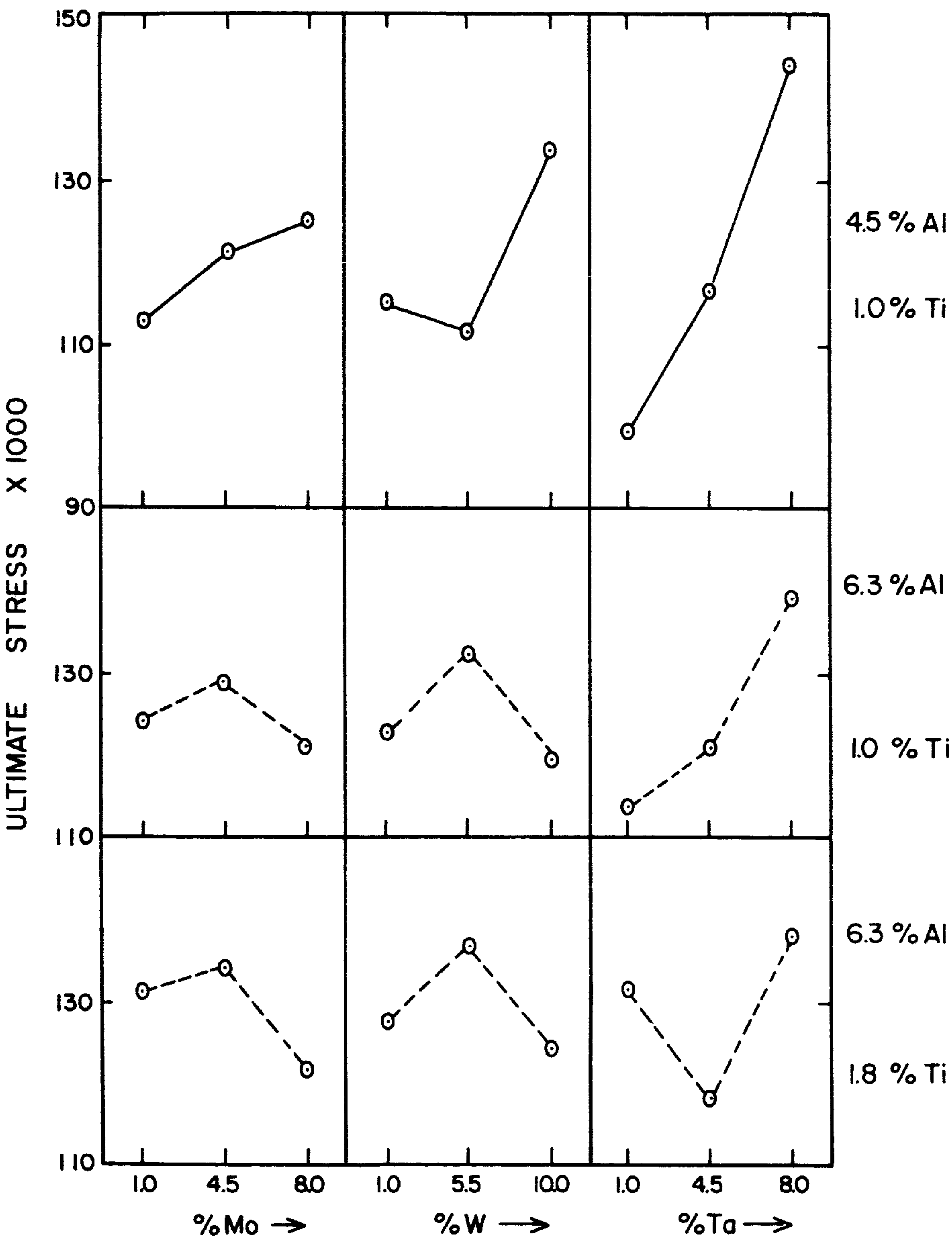


Figure 14. Average 1400°F Tensile Strength for Nickel-Base Superalloys Containing Various Molybdenum, Tungsten, and Tantalum Levels and Having Aluminum and Titanium Contents Shown. (Solid Lines are Statistically Significant.)

9866

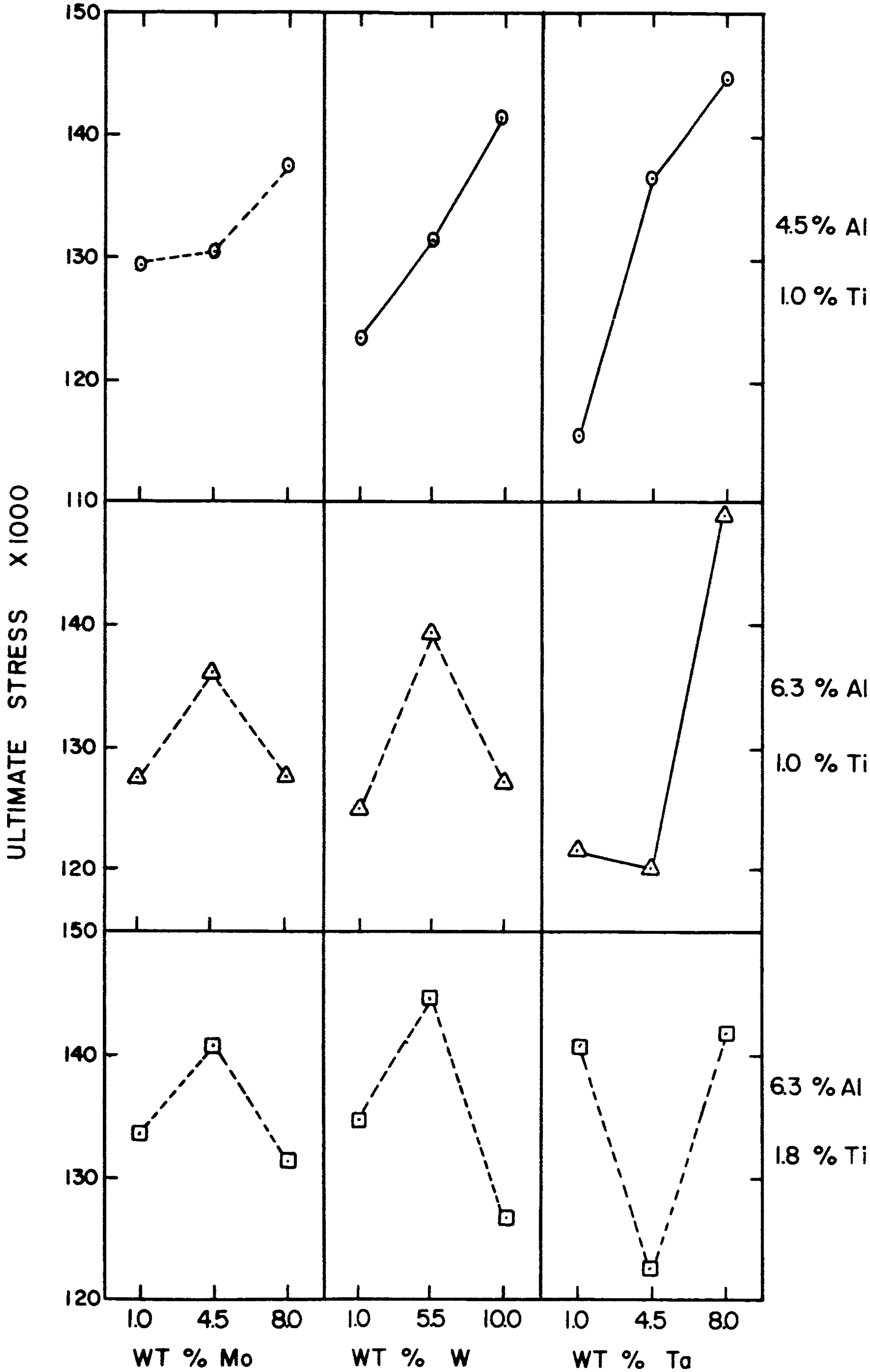
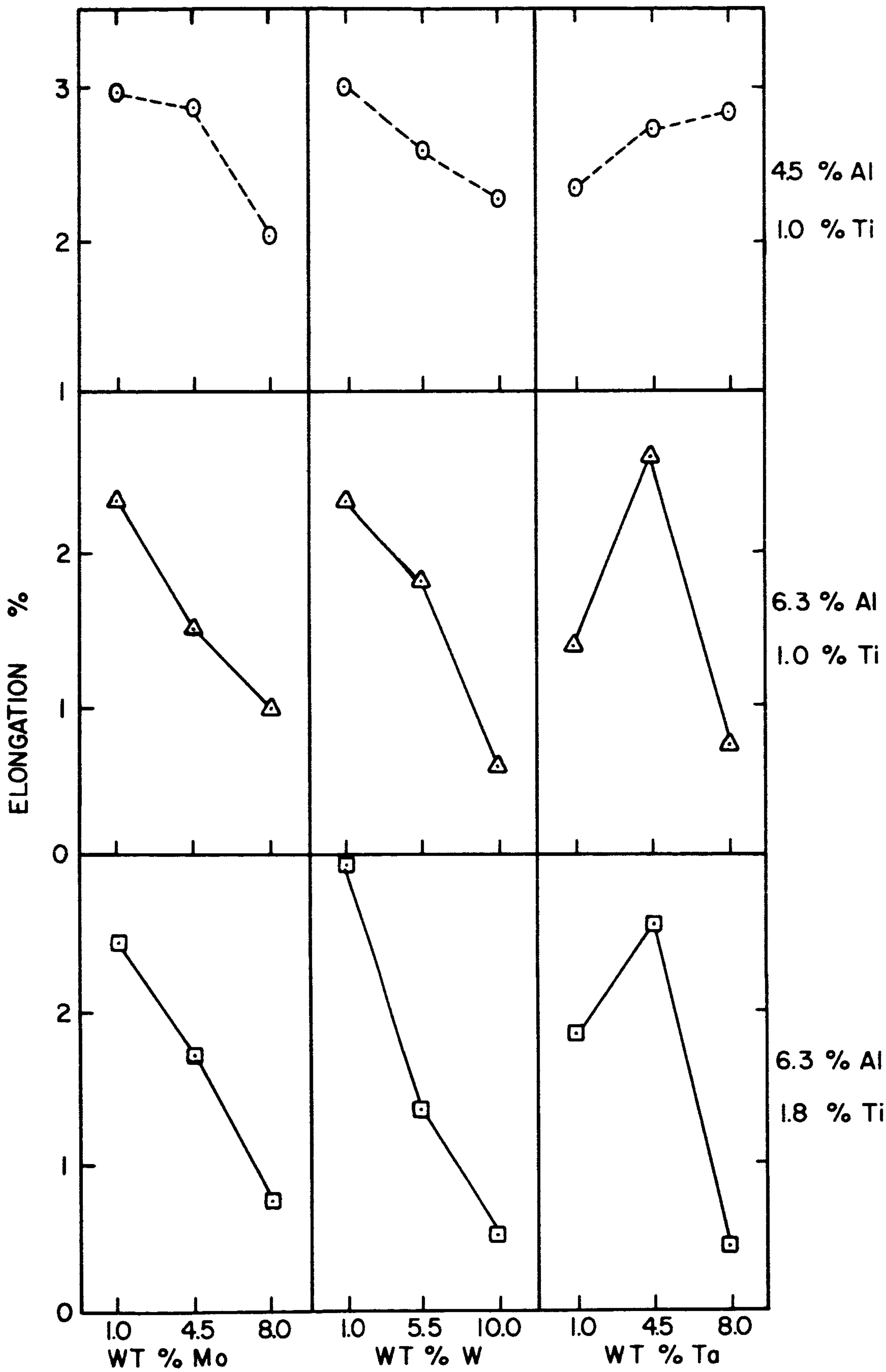


Figure 15. Average Room Temperature Tensile Strength for Nickel-Base Superalloys Containing Various Molybdenum, Tungsten, and Tantalum Levels and Having Aluminum and Titanium Contents Shown. (Solid Lines are Statistically Significant.)

9875



9664

Figure 16. Average 1400°F Tensile Elongation for Nickel-Base Superalloys Containing Various Molybdenum, Tungsten, and Tantalum Levels and Having Aluminum and Titanium Contents Shown. (Solid Lines are Statistically Significant.)

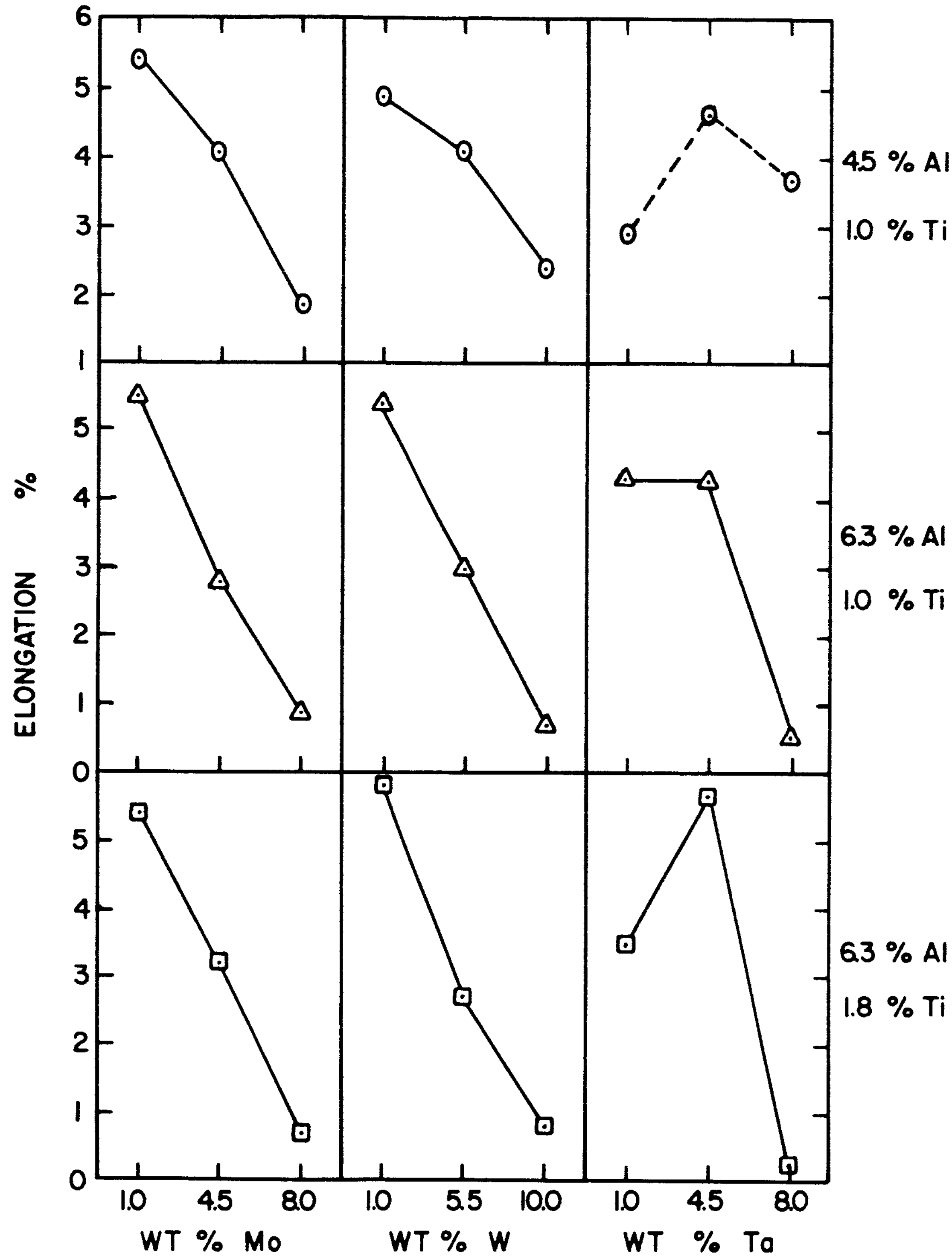
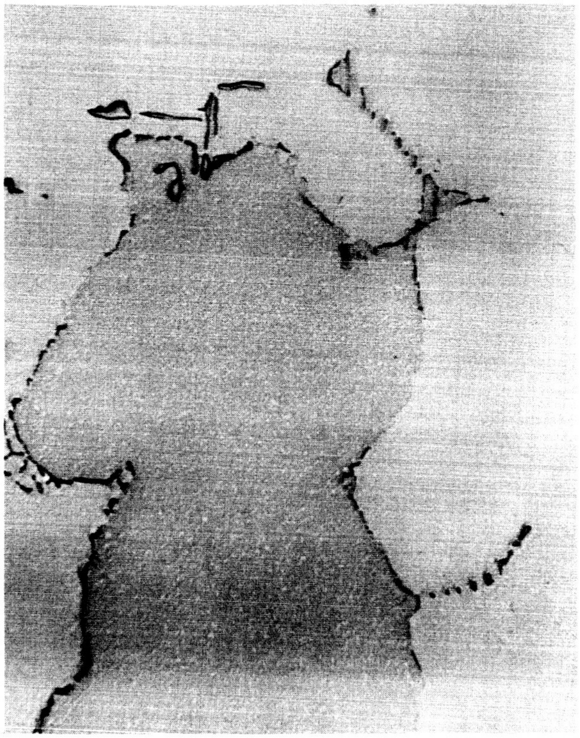


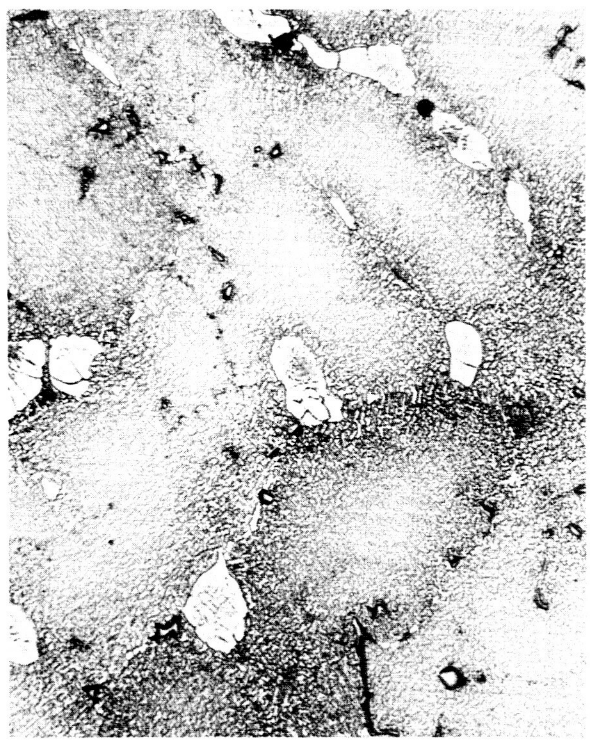
Figure 17. Average Room Temperature Tensile Elongation for Nickel-Base Superalloys Containing Various Molybdenum, Tungsten, and Tantalum Levels and Having Aluminum and Titanium Contents Shown. (Solid Lines are Statistically Significant.)



A. Alloy II, 250X

B. **Alloy II, 750X**

C. **Alloy III, 250X**



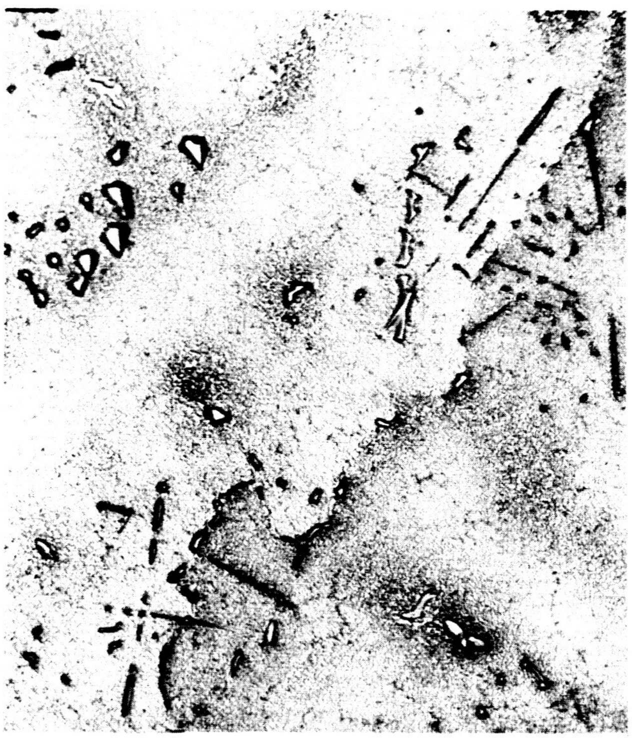
D. Alloy VI, 250X

Figure 18. Microstructure of Various Experimental Nickel-Base Superalloys Showing Carbide and Gamma Prime Formations in Lean Alloys (A, B, and C) and Massive Gamma Prime (White Areas) in Alloy VI (D). Etchant: 62% H_2O, 15% HF, 15% H_2SO_4, and 8% HNO_3.

9877

A. Alloy V, 250X

B. Alloy VII, 250X



C. Alloy XIII, 250X

D. Alloy XIII, 750X

Figure 19. Microstructure of Three Experimental Alloys Exhibiting Superior Stress Rupture Properties. Alloy V (A) is Primarily Gamma Prime Strengthened and Alloys VII and XIII (B, C, and D) are Heavily Solid Solution Strengthened.

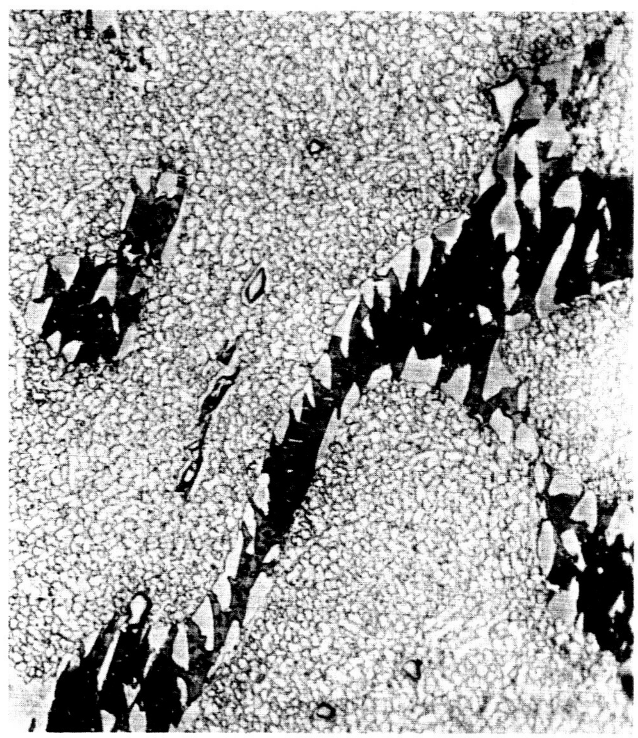
9874

Etchant: 62% H_2O, 15% HF, 15% H_2SO_4, and 8% HNO_3.

A. Unetched, 250X

B. Etched, 250X

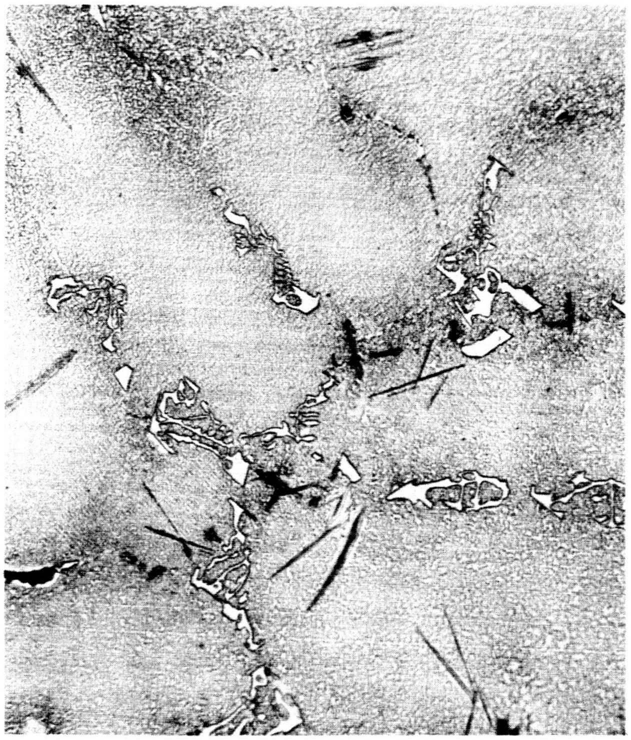
C. Lightly Etched, 750X



D. Etched, 750X

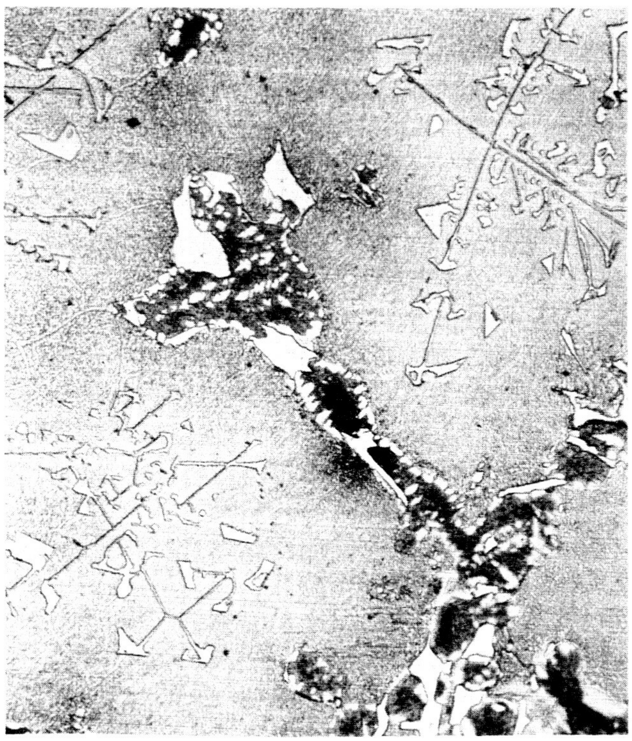
Figure 20. Microstructure of Alloy IX Containing a High Percentage of Refractory Metal Additions and 6.3%Al-1.8%Ti Showing Appearance of Dark Etching Phase Adjacent to Light Gamma Prime Phase. Etchant: 62%H_2O, 15%HF, 15%H_2SO_4, and 8%HNO_3.

9870



A. Alloy XVII Showing
Needlelike Phase

B. Alloy XXV, Containing
4.5%Al and 1.0%Ti

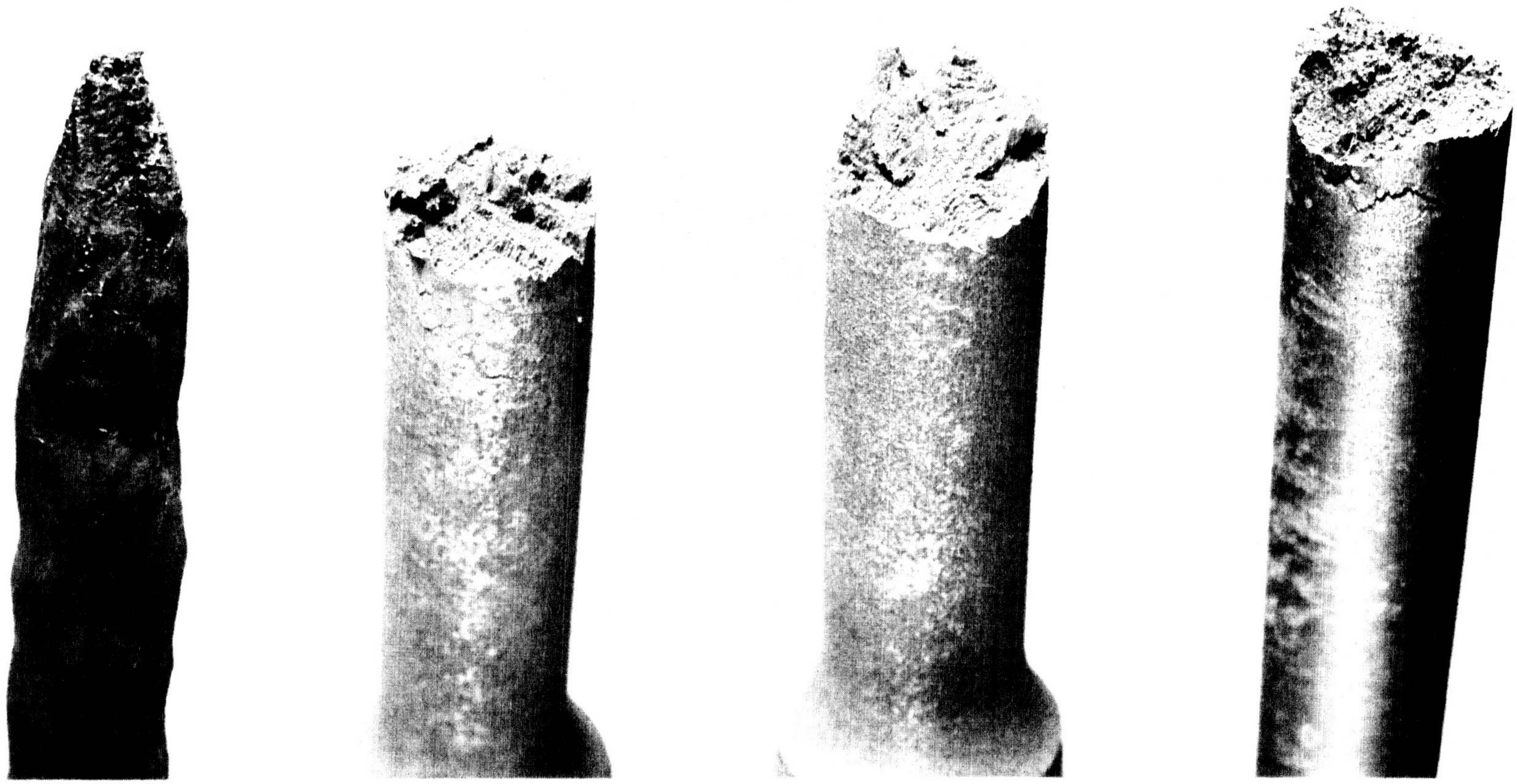


C. Alloy XXVI Containing
6.3%Al and 1.0%Ti

D. Alloy XXVII Containing
6.3%Al and 1.8%Ti

Figure 21. Microstructure of Various Experimental Nickel-Base Alloys Showing Appearance of Possible Sigma Phase (A) and Effect of Increasing Aluminum and Titanium Contents Upon Intermetallic Formation (B, C, and D). All 250X Magnification. Etchant: 62%H_2O, 15%HF, 15%H_2SO_4, 8%HNO_3.

9869

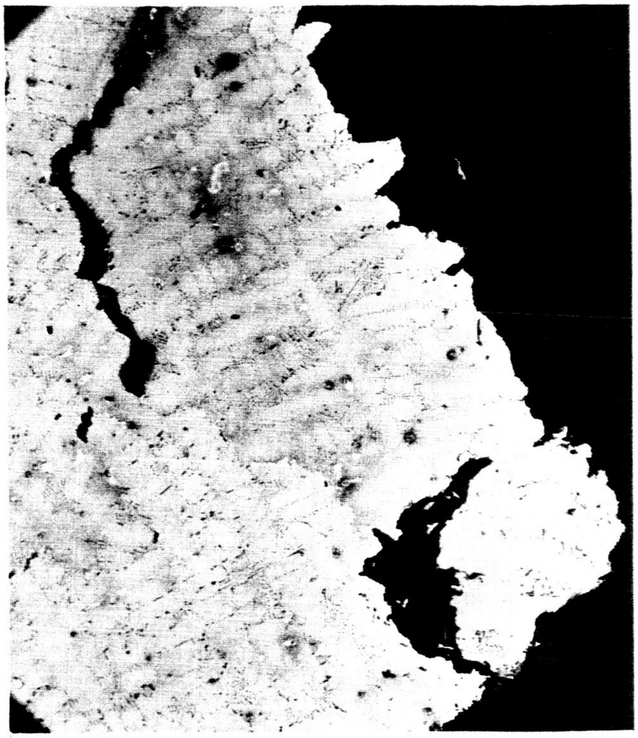


A. Alloy I B. Alloy V C. Alloy VII D. Alloy XXI

Figure 22. Appearance of Typical Stress Fractures in Experimental Nickel-Base Superalloys Showing Ductile (A) and Brittle (B, C, and D) Behavior. 5X Magnification

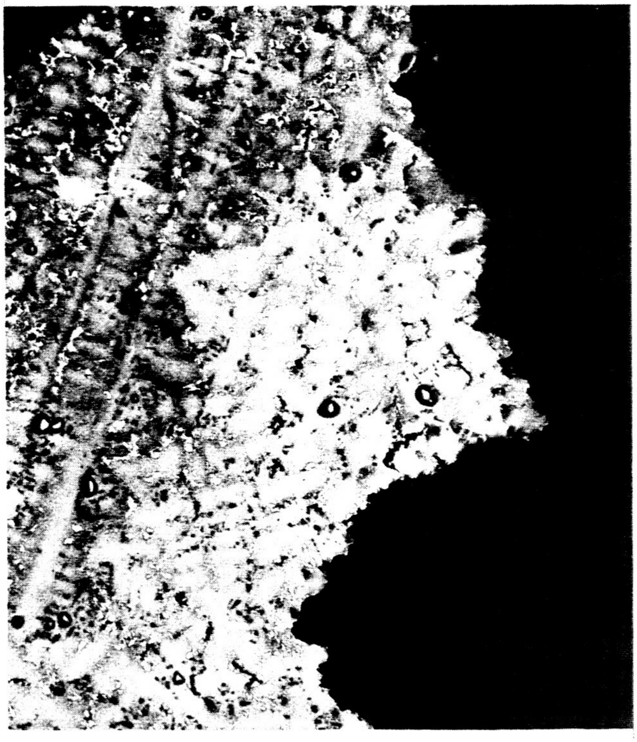
9876

A. Alloy I



B. Alloy VII

C. Alloy XVI



D. Alloy XXI

Figure 23. Microstructural Appearance of Various Nickel-Base Superalloys Adjacent to Fracture Showing Intragranular Nature of Stress Rupture Failure. 60X Magnification.
Etchant: 62% H_2O, 15% HF, 15% H_2SO_4, and 8% HNO_3.

9871

9872

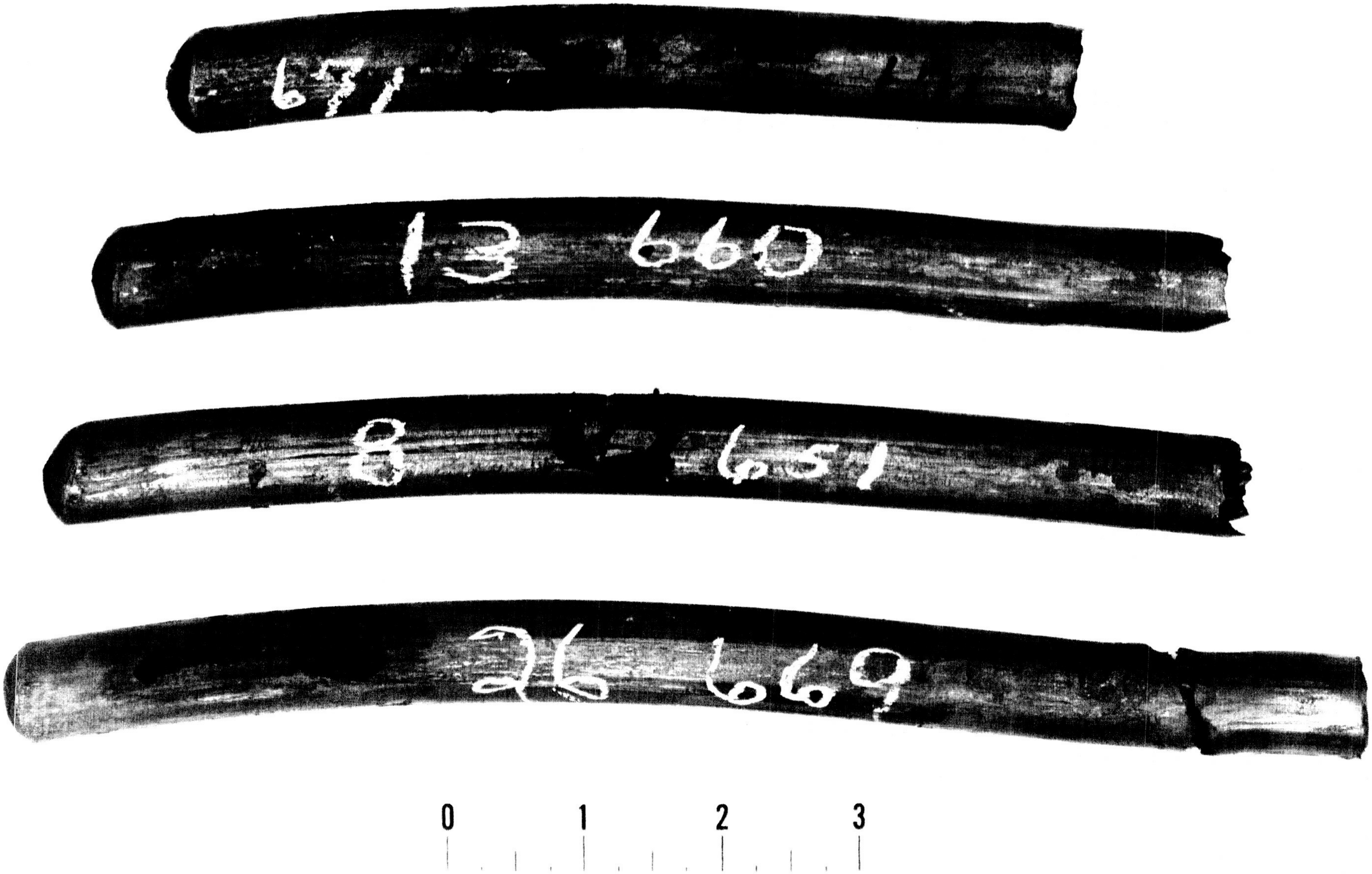


Figure 24. Typical Extrusions (4:1 Area Reduction) of Experimental Cast Nickel-Base Superalloys.

APPENDIX

Sample Statistical Calculation

See Table VIII for data (log 2000°F stress rupture life x 10^2 at 15,000 psi)

Series I base with 4.5% Al and 1.0T Ti

Summation of all values $\Sigma(X)$ = 51.7518 (27 values)

Molybdenum

	Average
Sum of all 1.0% Mo results = 13.8155 (9 values)	1.5351
Sum of all 5.5% Mo results = 18.8858 (9 values)	2.0984
Sum of all 8.0% Mo results = 19.0505 (9 values)	2.1167

Sum of Squares Calculation

$$\frac{(13.8155)^2}{9} + \frac{(18.8858)^2}{9} + \frac{(19.0505)^2}{9} - \frac{(51.7518)^2}{27} = 1.9682$$

Tungsten

	Average
Sum of all 1.0% W results = 12.3781 (9 values)	1.3753
Sum of all 5.5% W results = 17.1646 (9 values)	1.9072
Sum of all 10.0% W results= 22.2091 (9 values)	2.4677

Sum of Squares Calculation

$$\frac{(12.3781)^2}{9} + \frac{(17.1646)^2}{9} + \frac{(22.2091)^2}{9} - \frac{(51.7518)^2}{27} = 5.3706$$

Tantalum

	Average
Sum of all 1.0% Ta results = 10.2031 (9 values)	1.1337
Sum of all 4.5% Ta results = 14.8696 (9 values)	1.6522
Sum of all 8.0% Ta results = 26.6791 (9 values)	2.9643

Sum of Squares Calculation

$$\frac{(10.2031)^2}{9} + \frac{(14.8696)^2}{9} + \frac{(26.6791)^2}{9} - \frac{(51.7518)^2}{27} = 16.0258$$

Total

Sum of squares of all results $\Sigma(X^2)$ = 123.2497

Subtracting from this value the sum of all values $\frac{(\Sigma X)^2}{27}$ = 99.1944

Total = $\Sigma(X^2) - (\Sigma X)^2/27$ = 24.0553

Degrees of Freedom

Degrees of freedom for each Mo, W and Ta = N-1 = 2
(where N is the number of tests conducted)

Degrees of freedom of total = N-1 = 27-1=26

Residual and Interaction

Degrees of freedom = 26-6=20

Sum of squares of residual and interaction terms =

Total - (sum of other sums of squares)

$$24.0553 - (1.9682 + 5.3706 + 16.0258) = 0.6907$$

Mean Square

Mean square of Mo = sum of squared ÷ degrees of freedom
= 1.9682 ÷ 2 = 0.9841

Tabular Form

	Sum of Squares	Degrees of Freedom	Mean Square	F Ratio*
Mo	1.9682	2	0.9841	28.52
W	5.3706	2	2.6853	77.83
Ta	16.0258	2	8.0129	232.26
Residual and Interaction	0.6907	20	0.0345	
Total	24.0553	26		

*The F ratio = $\frac{\text{Mean Square of the Variable}}{\text{Mean Square of Residual and Interaction}}$

Consulting a table of F ratios,

F ratio (2,20) for 99.0% significance $\geq$ 5.85

Therefore, for this example all variables are considered significant.