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# EFFECTS OF CARBON AND HAFNIUM CONCENTRATIONS IN WROUGHT POWDER-METALLURGY SUPERALLOYS BASED ON NASA IIB-11 ALLOY

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16. Abstract

NASA IIB-11, a candidate alloy for advanced-temperature turbine engine disks, and four modifications of that alloy with various C and Hf concentrations were produced as cross-rolled disks from prealloyed powder that was hot isostatically compacted. The mechanical properties, microstructures, and phase relations of the alloys are discussed in terms of their C and Hf concentrations. A low-C and high-Hf modification of IIB-11 had the best balance of mechanical properties for service below about 750° C. However, because of their finer grain sizes, none of the powder-metallurgy alloys produced had the high-temperature rupture strength of conventionally cast and wrought IIB-11.

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# EFFECTS OF CARBON AND HAFNIUM CONCENTRATIONS IN WROUGHT POWDER-METALLURGY SUPERALLOYS BASED ON NASA IIB-11 ALLOY

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## SUMMARY

This program was undertaken to determine the effects of variations in carbon (C) and hafnium (Hf) concentrations on the properties, microstructure, and phase relations of an experimental wrought nickel-base superalloy, NASA IIB-11. This alloy is a candidate material for turbine disks that could operate with rim temperatures as high as $750^{\circ}$ C.

Alloy IIB-11 and four modifications of that alloy with various C and Hf concentrations were made into disks about 18 centimeters in diameter and 3 centimeters thick. The disks were produced by cross rolling billets of prealloyed powder that had been hot isostatically compacted.

The alloy modification with lower C and higher Hf concentrations than IIB-11 exhibited the best mechanical properties at temperatures of $760^{\circ}$ C and below and the best stability during long-time aging at $870^{\circ}$ C. This alloy had a rupture life of 230 hours at $760^{\circ}$ C and 620 $MN/m^2$ and an ultimate tensile strength of 1590 $MN/m^2$ at room temperature. These properties are at least equivalent to those of other candidate alloys for a $750^{\circ}$ C turbine disk on a strength-to-density basis.

All of the powder-metallurgy alloys investigated herein, however, had high-temperature rupture strengths that were more than 9 percent lower than that of cast and wrought IIB-11. These lower rupture strengths appeared to be due to failure to achieve as large a grain size in any powder-metallurgy alloy as had been produced in the cast and wrought IIB-11.

Grain growth was more difficult to achieve as Hf concentration was increased and, to a lesser extent, as C concentration was decreased among the alloys since these changes increased the gamma-prime solvus temperature. The amount of gamma prime in the alloys increased with decreasing C concentration but was not affected by Hf concentration. Increasing Hf did, however, promote cellular precipitation of the gamma prime. Also, increasing Hf, as well as increasing C, increased the tendency toward MC formation relative to that of $M_6C$ during exposure at $870^{\circ}$ C.

# INTRODUCTION

Alloy NASA IIB-11 was identified in a study of compositions intended to provide advanced wrought nickel-base superalloys. That work was conducted by Universal-Cyclops Specialty Steel Division under contract to the Lewis Research Center of NASA (ref. 1). In the form of 1.3-centimeter-diameter wrought bar produced from a cast ingot, IIB-11 exhibited greater high-temperature tensile and stress-rupture strengths than any wrought nickel-base superalloy known. Alloy IIB-11 is considered an excellent candidate for further development as a material for gas turbine disks that could operate with rim temperatures as high as 750° C. It is indicated in reference 1 that the interaction between carbon (C) and hafnium (Hf) concentrations was very important in determining strength in the alloys of that statistically designed experiment. These results suggested that still greater strength might be obtained from the IIB-11 composition by proper modification of the C and Hf concentrations.

Besides studying the effects of C and Hf concentrations on mechanical properties, it was desired to determine the effects of Hf on the microstructure and phase compositions of wrought alloys produced from prealloyed powder. Studies (refs. 2 to 5) have shown that Hf has a profound effect on the microstructure of cast nickel-base superalloys, but little was known about its effects in wrought superalloys. In particular it was expected that Hf, which forms a very stable MC carbide, might benefit superalloys to be produced from prealloyed powders. This effect was expected since increased carbide stability should lessen the tendency of some alloys to form carbide networks at the boundaries of the powder particles during consolidation. Therefore, this study was conducted to determine the effects of C and Hf concentrations on the mechanical properties, microstructures, and phase relations in alloys based on the NASA IIB-11 composition produced from prealloyed powder.

The alloys studied herein were produced from prealloyed powder since it was anticipated that any optimized alloy based on the IIB-11 composition would most likely be produced by powder-metallurgy methods. Large cast ingots of such a heavily alloyed composition would be considered difficult to work successfully because of segregation.

Alloy IIB-11 and four modifications of that alloy with C and Hf concentrations varying about those of alloy IIB-11 were produced. Disks about 18 centimeters in diameter and 3 centimeters thick were produced by cross rolling billets of prealloyed powders that had been hot isostatically compacted. The disks were cut into pieces that were variously heat treated and then tested in tension and stress rupture. Also, extensive studies were made of the microstructure and phase compositions of the alloys.

# MATERIALS AND PROCEDURES

## Materials

The materials used were alloy IIB-11 and four modifications of that alloy containing different C and Hf concentrations. The aim C and Hf concentrations for the five alloys are shown schematically in figure 1. All the alloys were produced from a single vacuum-melted heat of remelt stock. The aim composition for the remelt stock was that of the low-C and low-Hf modification (alloy 1), which is otherwise the same as that of IIB-11 (ref. 1). The aim composition and the vendor's analysis of the remelt stock are given in table I. The actual composition was very close to the aim.

Prealloyed powders of the five alloys were produced by the hydrogen atomization method described in reference 6. Alloy 1 powder was produced by direct atomization of the remelt stock. The other alloys were produced by adding C and/or Hf to the remelt stock during melting for atomization. Analyses by atomic absorption spectroscopy on the hot isostatically compacted and cross-rolled disks produced from the powders are presented in table I. The composition of each alloy was very close to that proposed. However, the Hf concentrations of alloys 2 and 4 (~2.5 wt %) and alloy 5 (1.4 wt %) are higher than the Hf concentrations desired (2 and 1 wt %, respectively).

The oxygen concentrations of the alloys with Hf added (97, 195, and 136 ppm for alloys 2, 4, and 5, respectively) were considerably higher than those of the essentially Hf-free alloys (~50 ppm). The additional oxygen in the alloys with Hf added was most likely due to an increased reaction with the oxide crucible material during melting for atomization.

The alloy powders were compacted by hot isostatic pressing. The powders were loaded in air into cans fabricated from 1.2-millimeter-thick, aluminum-killed 1010 steel. The cans were 14.8 centimeters in diameter and 8.5 centimeters high. They had been annealed under vacuum before filling, and they were vibrated during filling to achieve maximum density. After filling, all five cans were connected to a manifold and simultaneously evacuated by a single diffusion pump. During evacuation the cans were heated at about $500^\circ$ C; evacuation was continued for 16.5 hours. At the end of the evacuation cycle the pressure in the manifold was less than $10^{-3}$ $N/m^2$ ($10^{-5}$ torr). The cans were sealed under vacuum by crimping and then by welding shut the tubes provided for connecting the cans to the vacuum system. After sealing, the cans were externally pressurized in 1.4-$MN/m^2$ He and then checked with a He leak detector. No leaks were found.

The cans were hot isostatically pressed at 100 $MN/m^2$ and $1195^\circ \pm 5^\circ$ C for 2 hours, all in a single run.

Prior to cross rolling, the steel canning material was removed from the billets by pickling. About 3 millimeters was ground from all surfaces to remove any poorly consolidated material and to make the top and bottom faces flat and parallel. The dressed preforms were 12 centimeters in diameter and 6.5 centimeters high. These preforms were then canned for rolling in 1.3-centimeter-thick Unitemp HX sheet. The canned preforms were reduced about 60 percent in thickness by cross rolling at $1080^{\circ}\pm15^{\circ}$ C. The reductions were limited to 10 percent per pass, and the pieces were reheated 15 minutes between passes. In cross rolling, the pieces were turned $90^{\circ}$ between each pass so that they remained essentially round.

The thickness of the cross-rolled disks measured after removal of the canning material by pickling was about 2.8 centimeters except at the very edges. There the disks were about 3.0 centimeters thick. Thus, the disks had been reduced 57 percent except for the outer edges, which received about a 53 percent reduction. The final diameter of the disks was about 18 centimeters.

The appearance of the cross-rolled disks is shown in figure 2. None of the alloys except alloy 2 exhibited even a small edge crack. Alloy 2 had, however, three large cracks and about twice that number of small edge cracks. By careful cutting it was still possible, though, to obtain many sound mechanical test specimens.

Before the disks were cut, they were ultrasonically inspected. Except, of course, for alloy 2 all the disks were very free of defects; no defect larger than 4 millimeters was indicated. Later Zyglo inspection of the mechanical test specimens machined from alloy 2 showed that the uncracked areas were indeed free of smaller defects.

Pieces 6 centimeters by 1.2 centimeters were cut from the disks and then cut in half to make pieces 6 centimeters by 1.2 centimeters by 1.3 centimeters. These were then ground to dumbbell-shaped blanks for mechanical test specimens 0.66 centimeter in diameter in the center section. The reduced section and the threaded grips were not ground until after heat treatment. The axes of the specimens were parallel to the direction of the last rolling pass given to the disks.

## Heat Treatments

Each alloy was evaluated in three different heat-treated conditions that varied only in the initial solution treatment. Solution treatment temperatures in the range $1220^{\circ}$ to $1245^{\circ}$ C were chosen based on the results of metallographic examinations of bars of the alloys after they had been heated for 2 hours in a gradient furnace. Temperatures along these bars were measured with thermocouples strapped to both ends and the middle.

The same aging treatment, that used for cast and wrought IIB-11 (ref. 1), was employed for all the heat treatments; that is, $870^{\circ}$ C for 16 hours followed by $760^{\circ}$ C for 16 hours. The specimens were rapidly cooled in air by being placed under a fan both after the solution treatment and after the first aging step.

The specific solution treatments applied to each alloy are described, together with the resultant mechanical properties, in the section RESULTS AND DISCUSSION.

After heat treatment a few mechanical test specimen blanks were given 1500 hours of exposure at 870° C under argon atmosphere to test the stabilities of the alloys' mechanical properties and microstructures.

### Mechanical Testing

The specimens for tensile and stress-rupture testing had reduced sections 0.48 centimeter in diameter and 2.3 centimeters long. All tests were conducted in air and in accordance with American Society for Testing Materials (ASTM) recommended practice. Elongation was determined after fracture by dividing the specimen elongation measured from punch marks on the specimen shoulders by the length of the reduced section.

### Metallography and Phase Analysis Procedures

Metallographic examinations were performed on the alloys after each processing step. Photomicrographs of the as-hot-isostatically-pressed alloys were made by the company that cross rolled the disks. They used a surface replication technique. The other metallographic specimens were all etched with a solution of 33 parts each of nitric acid, acetic acid, and water plus 1 part of hydrofluoric acid.

The grain sizes of the alloys in the various heat-treated conditions were determined from the grip sections of the specimens that were tested in tension at room temperature. A lineal counting method was used.

The gamma-prime solvus temperature for each alloy was determined, from a gradient annealed bar, as the temperature corresponding to the position where the coarse gamma prime in the as-rolled alloy appeared completely solutioned.

Qualitative electrolytic phase extractions, using the methods of reference 7, were performed on alloy samples given the optimum heat treatment in order to determine the amounts and types of phases present.[1] The phases present in the extracted residues were determined by X-ray diffraction. The extracted residues were also analyzed chemically by the methods of reference 8. In order to determine the compositions of gamma and gamma-prime phases a mass-balance approach like that of Kriege and Baris (ref. 9) was used. The phases present in alloy samples that had been aged for

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[1] Much of this work was performed by Jere Lou Luttner, a summer student at Lewis.

1500 hours at 870° C were also determined by X-ray diffraction studies of electrolytic extractions.

## RESULTS AND DISCUSSION

### Microstructures of Alloys as Hot Isostatically Pressed

None of the alloys exhibited carbide precipitation at the prior powder-particle boundaries after hot isostatic pressing. Thus, the carbides in all of the alloys were sufficiently stable to prevent such dissolution and reprecipitation, and the opportunity to observe any increased stability of the MC carbides in the alloys containing Hf did not arise. The results of other related studies, which are discussed in the section Phase Analyses of Alloys Given Preferred Heat Treatment do show, however, that Hf increases the stability of the MC carbide.

### Microstructures of Alloys after Various Heat Treatments

It is helpful to discuss the microstructures of the alloys in the various conditions of heat treatment that were studied, particularly their grain sizes, before discussing the corresponding mechanical property data. Grain size is primarily determined by the relation of the solution treatment temperature to the alloy's gamma-prime solvus temperature. Solution treatment at a temperature somewhat above the alloys' solvus temperature is necessary to produce considerable grain growth.

The approximate gamma-prime solvus temperatures of the alloys are shown in figure 3 relative to their C and Hf concentrations. It can be seen that the solvus temperatures increase with increasing Hf concentration and to a lesser extent with decreasing C concentration among the alloys. Figure 4 shows, however, that while the amount of gamma prime in the alloys also increases with decreasing C concentration, it is not affected by Hf concentration. This fact is simply noted here. These and other results of the electrolytic phase extractions performed are discussed further in the section Phase Analyses of Alloys Given Preferred Heat Treatment.

The grain sizes of the alloys in the various heat-treated conditions that were studied are shown in figure 5 relative to their C and Hf concentrations. The heat treatments are identified by their solution treatments, the only difference among them. For each heat treatment the alloys exhibiting the largest grain sizes were those with the lowest gamma-prime solvus temperatures. Thus, the effect of solution temperature on grain size can be judged from figure 5; however, it is more readily seen in figure 6. Here it has been attempted to compensate for the alloys' differing gamma-prime solvus temperatures by showing the grain size of the alloys relative to the solution treatment temperature minus the alloy's gamma-prime solvus temperature. As the solution treatment temperature approached the solvus temperature, grain size began to increase at a higher rate.

The largest average grain size developed was in alloy 3 (high C and low Hf) by using 1245° C solution treatment. This grain size was 70 micrometers, or about ASTM grain size 5. The cast and wrought IIB-11 (ref. 1) developed a 90-micrometer grain size upon solution treatment at 1230° C for 2 hours (private communication with W. B. Kent, Universal-Cyclops, Bridgeville, Pa.).

When substantial grain growth was produced in any of these alloys, their appearance suggested a somewhat abnormal growth character. This was most pronounced in alloy 3, which exhibited a range of grain diameters differing by a factor of about 10.

The heat treatment that involved a solution treatment of 1220° C for 4 hours was not applied to alloy 2 since examination of the bar annealed in the gradient furnace showed that this temperature would yield only a very fine grain size. Instead, some specimens of alloy 2 were heat treated by using a 1230° C/4 hour solution treatment. Solution treatments above 1245° C were avoided, even though the gamma-prime solvus of alloy 2 is about 1260° C. Examination of the gradient annealed bar showed higher solution temperatures would result in a considerable amount of gamma prime being precipitated in a coarse cellular morphology upon cooling. Cellular precipitation has been observed in other Hf-bearing alloys produced from prealloyed powders that have been studied here and by others (private communication with J. M. Larson, International Nickel Co., Suffern, N. Y.).

The microstructures of alloys 2 and 5 given the heat treatment involving a 1230° C/ 2 hour solution treatment are shown in figure 7. As is explained in the following section, this heat treatment was the most preferred, of those investigated, based on the mechanical properties developed. Even after this solution treatment there were some areas of cellular gamma prime in alloy 2. The appearance of alloy 5 was more representative of the other alloys.

## Mechanical Properties of Alloys after Various Heat Treatments

In this section all of the alloys are considered more or less as a group in order to describe the effects of the various heat treatments on mechanical properties. In the following section the effects of the alloys' C and Hf concentrations are discussed in more detail, but only for the heat treatment involving the 1230° C/2 hour solution treatment. However, all of the mechanical property data obtained in this study are presented for record in two tables: stress-rupture data in table II, and tensile data in table III. Data for cast and wrought IIB-11 (ref. 1) are also presented for comparison.

The 1230° C solution treatment was preferred over both those at 1220° and 1245° C for several reasons. The fine grain sizes produced in all the alloys, except alloy 3, by the 1220° C solution treatment led to stress-rupture lives at 760° C and 620 $MN/m^2$ that were almost an order of magnitude lower than those of the specimens solution treated at 1230° C. Also, in the 760° C tests the rupture elongations of all the alloys, including alloy 3, were less than about 1 percent when the 1220° C solution treatment was employed.

Use of the higher solution treatment temperature (1245° C) produced some general increase in grain size among all of the alloys relative to use of the 1230° C solution treatment. However, stress-rupture lives at 760° C and 620 $MN/m^2$ were generally decreased, and little if any improvement was shown at 980° C and 170 $MN/m^2$. Also, particularly for the Hf-containing alloys, use of the higher temperature solution treatment decreased the room-temperature 0.2 percent yield stress.

### Mechanical Properties of Alloys Given Preferred Heat Treatment

The stress-rupture and room-temperature tensile properties of those alloys that were given the preferred heat treatment, 1230° C/2 hour/rapidly air cooled + 870° C/16 hour/rapidly air cooled + 760° C/16 hour/air cooled, are discussed in detail in the following sections. Comparisons are made among the alloys and with the properties of cast and wrought IIB-11 (ref. 1).

Stress-rupture properties. - A Larson-Miller plot of the stress-rupture data is shown in figure 8, together with the data on cast and wrought IIB-11 from reference 1. A value of 20 has been assumed for the constant. Note first that the powder-metallurgy (PM) alloys tested here all have lower rupture strengths than cast and wrought IIB-11, where data exist for the cast and wrought IIB-11. In terms of rupture life, shown in table II, the PM alloys with the longest lives at 760° C and 620 $MN/m^2$ and at 980° C and 170 $MN/m^2$ had lives only about half those of the cast and wrought IIB-11. The PM alloy of the IIB-11 composition (alloy 5) had 9 to 26 percent lower rupture strength than cast and wrought IIB-11 over the range of Larson-Miller parameters where comparison can be made (fig. 8).

The principal reason for the higher rupture strength of the cast and wrought IIB-11 appears to be its larger grain size (~ 90 $\mu$m). Figure 8 shows that among all the alloys, including cast and wrought IIB-11, there is a good correlation between increasing grain size and increasing rupture strength at the higher values of the Larson-Miller parameter.

The effects of the C and Hf concentrations in the PM alloys are shown in figure 9. Refer again to the effects of the C and Hf concentrations on the grain size of the alloys (fig. 5) and on the percentages of gamma prime in the alloys (fig. 4).

At 650° C and 1040 $MN/m^2$ the alloys with the highest Hf concentrations (and smallest grain sizes) had the longest lives. Alloy 2 (low C and high Hf) had the longest average life (~800 hr). Alloy 4 (high C and high Hf) had the second longest life (~430 hr). The lower life of alloy 4 might be explained by its lower amount of gamma prime (51 wt % compared to 57 wt % for alloy 2). However, the same difference in the amounts of gamma prime exists between the two low-Hf alloys, and their rupture lives are both about the same (~200 hr). Alloy 5, the IIB-11 composition, which has the median Hf concentration and grain size of the five alloys, had the lowest life (~140 hr).

At 760° C and 620 $MN/m^2$ the stress-rupture lives of the alloys were most nearly the same. The larger grain size of the alloys with the lower Hf concentrations apparently becomes more important at 760° C than it had been at 650° C.

At the still higher temperature of 980° C and a stress of 170 $MN/m^2$, the effect of grain size finally dominates the stress-rupture lives of the alloys. At this temperature, increasing stress-rupture life among the alloys correlates well with their increasing grain size.

Figure 9 also shows that the rupture elongation of alloy 2 (low C and high Hf) and, to a lesser extent, its reduction in area were superior for all of the test conditions employed. The rupture elongation and the reduction of area of alloy 2 were also superior to those of cast and wrought IIB-11, particularly at 760° C and 620 $MN/m^2$ (table II).

Room-temperature tensile properties. - All tensile data are given in table III. The data for the materials being discussed here, those given the 1230° C solution treatment, show that all of these PM alloys have about equal or higher average yield strengths and ultimate tensile strengths than cast and wrought IIB-11. Alloy 2, the strongest of the PM alloys, has 1250 and 1590 $MN/m^2$ yield and ultimate strengths, respectively, compared to 1190 and 1470 $MN/m^2$ for the cast and wrought IIB-11. Both materials have about the same elongation (10 percent), but the cast and wrought IIB-11 may have more reduction of area (~13.5 percent as compared to 11.5 percent).

Average room-temperature tensile data for the five alloys are shown in figure 10 relative to their C and Hf concentrations. Yield strength appears to have benefited by increasing Hf concentration and to a lesser extent by decreasing C concentration, which increases the gamma-prime content of the alloys. The ultimate tensile strengths, elongations, and reductions in area show a complex variation with C and Hf concentration that is difficult to understand. Note that alloy 4 (high C and high Hf) and alloy 5 (the IIB-11 composition) are the poorest of the alloys. This was also true of the specimens given the other two solution treatments. The poor properties of these two alloys might be due to their high oxygen concentrations, 0.0195 and 0.0136 weight percent, respectively, for alloys 4 and 5. However, alloy 2 (low C and high Hf) also had a high oxygen concentration (0.0095 wt %), and it has the best room-temperature tensile properties of all five alloys. Thus, it appears that oxygen content alone cannot account for the poor properties of alloys 4 and 5.

## Room-Temperature Tensile Properties and Microstructures after Long-Time Exposure

Room-temperature tensile tests were performed on specimens of the alloys that were exposed for 1500 hours at 870° C after having originally received the heat treatment involving the 1230° C/2 hour solution treatment. This exposure had been applied to the cast and wrought IIB-11 discussed in reference 1, and a considerable reduction in tensile properties had resulted. It was also shown in reference 1 that exposure produced precipitation of a mu-phase in an acicular morphology and large "pools" of gamma prime along the grain boundaries, particularly at their intersections.

The tensile test results for the exposed samples (870° C for 1500 hr), including those for cast and wrought IIB-11, are presented in table III. Average tensile properties after exposure are shown in table IV. Data for cast and wrought IIB-11 (ref. 1) are also presented in these tables. The percentage reductions in the average tensile properties of the alloys because of exposure are shown in table V.

Alloy 2 (low C and high Hf), which had the best tensile properties before exposure, suffered the least from exposure. In particular, alloy 2 suffered much less loss in ductility than the other PM alloys or cast and wrought IIB-11. Alloy 4 (high C and high Hf) and alloy 5 (IIB-11), which had the lowest tensile properties before exposure, suffered the most from exposure. Also the PM IIB-11 (alloy 5) suffered the same or larger percentage reduction in all properties as for cast and wrought IIB-11.

Studies were also made of the phases present in the alloys and the microstructures of the alloys. Figure 11 shows the minor phases in the exposed alloys that were determined by X-ray diffraction studies of the electrolytic extractions. A small amount of mu-phase with lattice parameters $a = 4.750\times10^{-10}$ m and $c = 25.69\times10^{-10}$ m was clearly identified in the extract from alloy 2. The other alloys, except alloy 1 (low C and low Hf), also appeared to contain small amounts of mu-phase, although the identifications were not so certain as for alloy 2.

As is discussed in the section Phase Analyses of Alloys Given Preferred Heat Treatment, all of the alloys contained an MC-type carbide and somewhat less of an $M_6C$-type before exposure. After exposure, all of the alloys contained little if any $M_6C$, except alloy 1, which contained only $M_6C$. It is possibly because of the stability of the $M_6C$ in alloy 1 that it did not form the mu-phase as did the other alloys. Decomposition of $M_6C$, which occurred in the other alloys, would be expected to release molybdenum and tungsten, which would contribute to mu-phase formation.

Formation of the mu-phase does not appear to correlate, however, with the reduction in tensile properties. Alloy 2, which formed the most mu-phase, suffered the least reduction in tensile properties from exposure. The most obvious difference in the microstructures of alloy 2 and the others after aging is that a grain boundary film of gamma prime developed in all of the alloys except alloy 2. Figure 12 shows the difference in the microstructures of alloys 2 and 5. Alloy 5 exhibited the greatest development of the gamma-prime grain boundary film.

Note also in figure 12 that the mu-phase in alloy 2 did not precipitate in the acicular morphology seen in the cast and wrought IIB-11 (ref. 1). In fact, no acicular precipitate was observed in any of the exposed alloys.

## Overall Comparison of Mechanical Properties

Alloy 2 (low C and high Hf) is the best of the alloys produced in this study, considering at once their tensile strengths, tensile and rupture ductilities, and stabilities. Also, as shown in figure 8, the rupture strength of alloy 2 is high at the lower values of the time-temperature parameter. However, the small grain size developed in alloy 2 relative to those of the alloys with lower Hf concentration would appear to constitute a severe penalty at the higher values of the time-temperature parameter. Extrapolations to 10 000-hour lives, though they are unlikely to be very accurate, indicate that the allowable stresses for alloy 2 with this small grain size would be increasingly low relative to those for some of the other alloys as temperatures increase above $700^{\circ}$ C.

It is shown, however, in the following section that, even at the higher values of the Larson-Miller parameter, alloy 2 is as strong as other candidate alloys for a $750^{\circ}$ C gas turbine disk that have been reported in the literature.

## Comparison of Alloy 2 to Other Alloys

Figures 13 and 14 compare the stress-rupture strength and tensile properties, respectively, of alloy 2 given the preferred heat treatment with cast and wrought IIB-11 (ref. 1) and with AF115 and AF2-1DA produced by powder-metallurgy methods. The latter two are also potential materials for gas turbine disks that would operate with rim temperatures as high as $750^{\circ}$ C. Data for AF115 and AF2-1DA are taken from references 10 and 11, respectively. Stresses for each alloy have been compensated for their differing densities by multiplying by the density of alloy 2 (8.80 g/cm$^3$) divided by the density of the particular alloy. This in effect increases the apparent strengths of AF115 and AF2-1DA, which have densities of 8.36 and 8.28 g/cm$^3$, respectively.

Cast and wrought IIB-11 clearly has a considerable advantage in rupture strength over the other alloys, about 20 percent higher at a Larson-Miller parameter of 42.3 and probably even higher at higher parameter values. Alloy 2 and AF115 have about equivalent rupture strengths, and that of AF2-1DA is somewhat lower.

Since tensile properties were measured only at room temperature in this study, comparisons with the other alloys can be made only at that temperature. It can be seen that the room-temperature tensile properties of alloy 2 and AF115 are about equivalent on this density-corrected basis. Compared to these two alloys, cast and wrought IIB-11 has a little lower ultimate strength (~ 7 percent lower), and AF2-1DA has both lower ultimate and 0.2-percent-yield strengths (both about 14 percent lower). The elongations of the alloys do not appear to differ greatly.

## Phase Analyses of Alloys Given Preferred Heat Treatment

The differences in C and Hf concentrations among the alloys affected the amounts of the gamma-prime and carbide phases formed, the compositions of the gamma and gamma-prime phases, and particularly the compositions of the carbides.

The amounts and types of carbides extracted electrolytically from the alloys are presented in table VI, together with the amounts of gamma prime that were determined. It may be seen that the total weight percentage of carbides in the alloys increases with increasing C concentration, as would be expected. X-ray diffraction studies of the extracted carbides showed that all of the alloys contained both MC- and $M_6C$-type carbides. It was not possible, though, to determine any change in the relative amounts of MC and $M_6C$ among the alloys by comparison of the relative intensities of their diffraction patterns. For both carbides the relative intensities of the individual reflections changed from alloy to alloy, and there were interferences between reflections from the two carbides that also changed from alloy to alloy. However, some reasonable suppositions can be made about changes in the relative amounts of the two carbides from chemical analyses of the extracted carbides.

The compositions of the carbides, MC and $M_6C$ together, extracted from the alloys are listed in table VII. The concentrations of several selected elements are shown in figure 15 relative to the C and Hf concentrations of the alloys. It may be seen that, as the C and Hf concentrations increase among the alloys, the amounts of those elements expected to be concentrated in the MC, namely Ti, Ta, and Hf, increase. Conversely, it may be seen that the amounts of those elements expected to be concentrated in the $M_6C$, namely Mo, W, and possibly Cr, increase with decreasing C and Hf concentrations among the alloys.

The conclusion that increased C and Hf concentrations favor the formation of MC relative to $M_6C$ seems reasonable. Increased C should move the equilibrium closer to the higher carbide (MC), and Hf is known to form a very stable MC carbide. This conclusion is also consistent with the observations presented previously about the relative stabilities of the MC and $M_6C$ carbides in the different alloys during long-time exposure. After 1500 hours at 870° C only $M_6C$ was found in the alloy with low C and low Hf (alloy 1); in the alloys with higher C and/or higher Hf, MC (but little or no $M_6C$) was found.

The compositions of the gamma and gamma-prime phases in the alloys are also affected by their C and Hf concentrations, though more subtly than those of the carbides. Tables VIII and IX show the calculated compositions of the gamma-prime and gamma phases, respectively. Besides data on the carbides, figure 15 shows the concentrations of selected elements in the gamma-prime and gamma phases of the alloys relative to their C and Hf concentrations. Also presented in figure 15 is the ratio of each element's concentration, in atomic percent, in the gamma-prime phase to that in the gamma phase.

The Hf concentration increases at about the same rate in both the gamma-prime and gamma phases as the Hf concentration of the alloy increases. Hf does not partition strongly to either phase in these alloys.

The Ta concentration increases in the gamma-prime phase with decreasing C and Hf in the alloy and exhibits the opposite behavior in the gamma phase. The ratio of Ta in the gamma-prime phase to that in the gamma phase varies greatly, from 7.8 in alloy 1 (low C and low Hf) to 1.4 in alloy 4 (high C and high Hf). Kriege and Baris (ref. 9) studied only one alloy containing Ta (i.e., B-1900), and they report much stronger partitioning to the gamma-prime phase (>20:1). Partitioning of Ta calculated from the data of Mihalison (ref. 5) is more like that found in this study and shows the same decrease in partitioning of Ta to the gamma-prime phase in Hf-bearing alloys. In IN-738 and IN-792, Ta concentrations were 6 to 7 times higher in the gamma-prime phase than in the gamma phase but were about equal in IN-792 with 1 weight percent Hf added. The calculated partitioning ratio for Hf (4.5) is, however, 3 to 4 times higher than those determined in the IIB-11 modifications.

The Mo, W, and Cr concentrations do not vary much in the gamma-prime or gamma phases except for those of alloy 2 (low C and high Hf). Alloy 2 has the lowest concentrations of these elements in the gamma-prime phase and the highest in the gamma phase. The partitioning ratios for Mo, W, and Cr in these alloys all lie within the range of values determined by Kriege and Baris. However, for alloy 2 the partitioning ratios for these elements all lie at the very bottom of the range of Kriege and Baris's values.

The elements in the alloys that have not been discussed did not vary in concentration in the individual phases to any large extent and will not be discussed. However, all the compositional data are presented in tables VII to IX.

## CONCLUDING REMARKS

The properties developed in the low C - high Hf powder-metallurgy modification of NASA IIB-11 (alloy 2) are competitive with those of other advanced high-temperature wrought alloys, however, the potential for still higher rupture strength, as shown by cast and wrought IIB-11, is a considerable incentive for further development work. Further compositional changes might be beneficial. However, it is difficult to predict from the results of this study that further small compositional modifications alone could yield the rupture strength shown by cast and wrought IIB-11. It appears clear that, in any further development work, a considerable portion of the effort must be devoted to changes in mechanical processing designed to yield a larger grain size in the powder-metallurgy alloy during heat treatment, one that is equivalent to the grain size of cast and wrought IIB-11.

Another reason for conducting further work on mechanical processing is that the working practice employed herein produced cracking in alloy 2 and might be expected to do so in other low C - high Hf modifications. It is believed likely that a little experimentation with the working schedule, possibly including heat treatment prior to working, would yield sound pieces. Alloy 2 after heat treatment was one of the most ductile of the five alloys in the tensile and rupture tests performed.

A further compositional modification to alloy 2 that might be beneficial from extrapolating the results of this study would be a still lower C concentration. Also, it is hoped that the Hf concentration could be lowered from 2.5 weight percent because of the element's cost and because it would decrease the gamma-prime solvus temperature, making it easier to achieve grain growth. And finally, though it appears that mu-phase formation in alloy 2 may not be very detrimental, an attempt should probably be made to make the alloy more stable by small reductions in the refractory element concentrations.

## SUMMARY OF RESULTS AND CONCLUSIONS

A study was undertaken to determine the effects of carbon (C) and hafnium (Hf) concentrations on the mechanical properties, microstructures, and phase relations in alloys based on NASA IIB-11 produced from prealloyed powder. The following results and conclusions were obtained:

1. At temperatures of 760° C and below the low C - high Hf modification (alloy 2) had the best balance of mechanical properties. It had average stress-rupture lives of 800 and 230 hours at 650° C and 1040 MN/m$^2$ and 760° C and 620 MN/m$^2$, respectively. Also, alloy 2 exhibited a room-temperature ultimate tensile strength of 1590 MN/m$^2$.

2. In stress-rupture tests at 980° C and 170 $MN/m^2$, rupture life increased with increasing grain size among the alloys investigated. In turn, grain size increased with increasing C and decreasing Hf concentrations.

3. The largest grain size produced in the powder-metallurgy alloys, by using a solution treatment of 1230° C for 2 hours, was 50 micrometers compared to 90 micrometers for cast and wrought IIB-11.

4. All of the powder-metallurgy IIB-11-based alloys had stress-rupture strengths more than 9 percent lower than that of cast and wrought IIB-11. For example, alloy 5, which had the IIB-11 composition, had rupture strength about 25 percent lower than that of cast and wrought IIB-11 at the highest Larson-Miller parameter where comparison could be made.

5. All of the powder-metallurgy IIB-11 alloys had room-temperature ultimate tensile strengths about equivalent to or higher than that of cast and wrought IIB-11 (1470 $MN/m^2$). Alloy 2 had a strength of 1590 $MN/m^2$.

6. On a strength-to-density basis, the mechanical properties of alloy 2 are equivalent to or higher than those of other alloys produced from prealloyed powder that might be used for turbine disks to operate with 750° C rim temperatures.

7. Alloy 2 exhibited the least degradation of room-temperature tensile properties after exposure for 1500 hours at 870° C, though it did form a mu-phase. The greater degradation of the other alloys appeared to be due to the formation of a nearly continuous film of gamma prime at their grain boundaries. This formation did not occur in alloy 2.

8. Grain growth is controlled primarily by dissolution of the gamma-prime phase. Increasing grain size among the powder-metallurgy alloys correlated with increasing difference between the solution treatment temperature and the alloys' gamma-prime solvus temperatures.

9. Increasing Hf concentration increased the gamma-prime solvus temperature, as did, to a lesser extent, decreasing C concentration.

10. The amount of gamma prime in the alloys appeared to be a function only of C concentration, increasing with decreasing C concentration.

11. Increasing Hf concentration promotes precipitation of gamma prime in a cellular morphology.

12. Increasing Hf and C concentrations promoted the stability of MC-type carbides relative to that of $M_6C$-type carbides during exposure at 870° C.

Lewis Research Center,
National Aeronautics and Space Administration,
Cleveland, Ohio, September 24, 1975,
505-01.

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TABLE I. - COMPOSITIONS OF REMELT STOCK AND ALLOY DISKS

| Element | Remelt stock | | Alloy[b] | | | | |
|---|---|---|---|---|---|---|---|
| | Aim | Actual[a] | 1 | 2 | 3 | 4 | 5 |
| | Concentration, wt % (except as noted) | | | | | | |
| C | 0.08 | 0.08 | 0.088 | 0.084 | 0.205 | 0.190 | 0.142 |
| Hf | ---- | <.01 | <.1 | 2.5 | <.1 | 2.6 | 1.4 |
| Al | 4.5 | 4.60 | 4.6 | 4.7 | 4.6 | 4.7 | 4.7 |
| Ti | .75 | .70 | .68 | .66 | .66 | .66 | .66 |
| Ta | 7.0 | 7.50 | 6.9 | 6.7 | 6.8 | 6.7 | 6.8 |
| W | 7.5 | 7.65 | 7.4 | 7.3 | 7.4 | 7.3 | 7.3 |
| Mo | 2.0 | 1.93 | 1.8 | 1.8 | 1.8 | 1.8 | 1.8 |
| V | .5 | .61 | .62 | .58 | .61 | .59 | .60 |
| Cr | 9.0 | 8.82 | 8.8 | 8.6 | 9.0 | 8.7 | 8.9 |
| Co | 9.0 | 9.10 | 9.5 | 9.1 | 9.4 | 9.2 | 9.2 |
| B | .02 | .019 | .02 | .02 | .02 | .02 | .02 |
| Zr | .1 | .07 | .08 | .09 | .07 | .07 | .07 |
| Fe | ---- | .33 | ------ | ------ | ------ | ------- | ------- |
| Mn | ---- | <.05 | ------ | ------ | ------ | ------- | ------- |
| Si | ---- | <.1 | ------ | ------ | ------ | ------- | ------- |
| P | ---- | <.01 | ------ | ------ | ------ | ------- | ------- |
| S | ---- | <.004 | ------ | ------ | ------ | ------- | ------- |
| Cu | ---- | <.1 | ------ | ------ | ------ | ------- | ------- |
| Pb | ---- | <1 ppm | ------ | ------ | ------ | ------- | ------- |
| Bi | ---- | <.5 ppm | ------ | ------ | ------ | ------- | ------- |
| O[c] | ---- | 14 ppm | 47 ppm | 97 ppm | 51 ppm | 195 ppm | 136 ppm |
| Ni | Bal. | Bal. | Bal. | Bal. | Bal. | Bal. | Bal. |

[a] Vendor's analysis.

[b] From atomic absorption spectroscopy.

[c] From analysis of the powders.

TABLE II. - STRESS-RUPTURE PROPERTIES OF ALLOYS

IN VARIOUS HEAT-TREATED CONDITIONS

| Heat treatment[a] | Test conditions | | Alloy specimen | Life, hr | Elongation, percent | Reduction of area, percent |
|---|---|---|---|---|---|---|
| | Temperature, °C | Pressure, MN/m$^2$ | | | | |
| 1230° C/2 hr/RAC<br>+870° C/16 hr/RAC<br>+760° C/16 hr/AC | 650 | 1040 | 1-1-1 | 138.2 | 2.1 | 4.2 |
| | | | 1-10 | 277.4 | 6.3 | 7.3 |
| | | | 2-8-1 | 814.6 | 7.9 | 8.9 |
| | | | 2-4 | 789.2 | 5.3 | 6.3 |
| | | | 3-1-1 | 251.7 | 4.7 | 6.3 |
| | | | 3-10 | 139.1 | 2.6 | 5.8 |
| | | | 4-1-1 | 564.8 | 6.8 | 7.8 |
| | | | 4-10 | 287.0 | 5.3 | 5.8 |
| | | | 5-1-1 | 95.3 | 3.7 | 5.2 |
| | | | 5-10 | 191.0 | 4.2 | 5.2 |
| | 760 | 620 | 1-2 | 187.1 | 2.1 | 4.7 |
| | | | 1-7-1 | 164.8 | 1.6 | 5.2 |
| | | | 1-11 | 234.8 | 2.1 | 4.2 |
| | | | 2-10 | 157.3 | 4.2 | 4.7 |
| | | | 2-11-1 | 301.7 | 4.7 | 6.2 |
| | | | 3-7-1 | 212.6 | 3.2 | 5.2 |
| | | | 3-11 | 119.6 | 1.1 | 2.6 |
| | | | 4-7-1 | 203.0 | 2.1 | 5.2 |
| | | | 4-11 | 111.9 | 2.1 | 4.7 |
| | | | 5-2 | 106.5 | 3.2 | 3.7 |
| | | | 5-7-1 | 333.6 | 2.1 | 4.2 |
| | | | 5-11 | 325.3 | 2.6 | 5.2 |
| | | | [b]IIB-11 | 511 | 2.2 | 4.2 |
| | 980 | 170 | 1-6 | 15.8 | 4.2 | 3.7 |
| | | | 2-7 | 3.0 | 6.3 | 10.3 |
| | | | 3-6 | 29.3 | 4.2 | 4.2 |
| | | | 4-6 | 6.6 | 3.2 | 7.8 |
| | | | 5-6 | 15.0 | 3.2 | 4.2 |
| | | | [b]IIB-11 | 63.0 | 5.4 | 5.9 |
| 1245° C/2 hr/RAC<br>+870° C/16 hr/RAC<br>+760° C/16 hr/AC | 760 | 620 | 1-2-1 | 104.0 | 1.1 | 3.2 |
| | | | 2-10-1 | 186.2 | 6.3 | 7.8 |
| | | | 3-2-1 | 161.8 | 3.2 | 5.8 |
| | | | 4-2-1 | 143.6 | 3.2 | 3.7 |
| | | | 5-2-1 | 171.0 | 2.1 | 4.2 |
| | 980 | 170 | 1-6-1 | 4.5 | 6.3 | 6.3 |
| | | | 1-13-1 | 17.3 | 5.3 | 4.7 |
| | | | 2-7-1 | 8.0 | 6.3 | 5.8 |
| | | | 3-6-1 | 14.7 | 4.2 | 2.6 |
| | | | 3-13-1 | 19.5 | 4.2 | 4.2 |
| | | | 4-6-1 | 10.3 | 6.3 | 6.8 |
| | | | 4-13-1 | 9.6 | 7.4 | 7.3 |
| | | | 5-6-1 | 20.3 | 4.2 | 4.7 |
| | | | 5-13-1 | 15.6 | 4.2 | 3.2 |
| 1220° C/4 hr/RAC<br>+870° C/16 hr/RAC<br>+760° C/16 hr/AC | 650 | 1040 | 1-10-3 | 173.9 | 5.3 | 5.3 |
| | | | 1-11-3 | 35.3 | 2.1 | 4.7 |
| | | | 3-10-3 | 166.5 | 5.3 | 6.3 |
| | | | 3-11-3 | (c) | (c) | (c) |
| | | | 4-10-3 | 64.9 | 2.1 | 4.7 |
| | | | 4-11-3 | 71.4 | 1.1 | 3.2 |
| | | | 5-10-3 | 41.7 | 1.6 | 2.6 |
| | | | 5-11-3 | 32.1 | 2.6 | 6.3 |
| | 760 | 620 | 1-3-3 | 46.1 | 0.8 | 3.2 |
| | | | 1-12-3 | 21.0 | ~0 | 1.0 |
| | | | 3-3-3 | 228.9 | 2.1 | 5.8 |
| | | | 3-12-3 | 207.9 | .5 | 3.2 |
| | | | 4-3-3 | 26.9 | .5 | 2.1 |
| | | | 4-12-3 | 49.5 | 1.1 | 1.6 |
| | | | 5-3-3 | 14.2 | .5 | 3.1 |
| | | | 5-12-3 | 20.6 | 1.1 | 2.1 |
| 1230° C/4 hr/RAC<br>+870° C/16 hr/RAC<br>+760° C/16 hr/AC | 650 | 1040 | 2-3-3 | 412.3 | 3.2 | 6.3 |
| | | | 2-6-4 | 785.0 | 5.8 | 6.2 |
| | 760 | 620 | 2-2-3 | 236.3 | 2.6 | 5.7 |
| | | | 2-6-3 | 195.3 | 1.6 | 2.6 |

[a] RAC = rapidly air cooled; AC = air cooled.

[b] Cast and wrought NASA IIB-11 (ref. 1).

[c] Defective specimen; metallographic examination showed that specimen had not been heat treated.

TABLE III. - TENSILE PROPERTIES OF ALLOYS IN VARIOUS HEAT-TREATED CONDITIONS

| Heat treatment[a] | Test temperature, °C | Alloy specimen | 0.2-Percent yield stress, MN/m$^2$ | Ultimate tensile strength, MN/m$^2$ | Elongation, percent | Reduction of area, percent |
|---|---|---|---|---|---|---|
| 1230° C/2 hr/RAC<br>+870° C/16 hr/RAC<br>+760° C/16 hr/AC | RT | 1-8 | 1210 | 1430 | 7.4 | 9.8 |
| | | 1-9 | 1180 | 1590 | 13.2 | 14.3 |
| | | 2-9 | 1200 | 1620 | 12.4 | 12.4 |
| | | 2-3 | 1290 | 1550 | 7.9 | 10.4 |
| | | 3-8 | 1130 | 1550 | 11.6 | 12.3 |
| | | 3-9 | 1220 | 1620 | 12.1 | 12.3 |
| | | 4-8 | 1170 | 1420 | 6.3 | 9.3 |
| | | 4-9 | 1280 | 1480 | 5.3 | 8.8 |
| | | 5-8 | 1190 | 1410 | 6.3 | 9.8 |
| | | 5-9 | 1230 | 1490 | 7.4 | 9.8 |
| | | [b]IIB-11 | 1190 | 1470 | 9.8 | 13.6 |
| 1245° C/2 hr/RAC<br>+870° C/16 hr/RAC<br>+760° C/16 hr/AC | RT | 1-8-1 | 1190 | 1530 | 10.5 | 12.8 |
| | | 2-9-1 | 1150 | 1600 | 14.5 | 14.5 |
| | | 3-8-1 | 1180 | 1540 | 10.5 | 10.8 |
| | | 4-8-1 | 1100 | 1480 | 10.5 | 12.3 |
| | | 5-8-1 | 1150 | 1440 | 8.4 | 10.9 |
| 1220° C/4 hr/RAC<br>+870° C/16 hr/RAC<br>+760° C/16 hr/AC | RT | 1-3-4 | 1190 | 1590 | 12.6 | 14.8 |
| | | 1-12-4 | 1270 | 1680 | 12.5 | 13.3 |
| | | 3-3-4 | 1200 | 1560 | 9.8 | 11.1 |
| | | 3-12-4 | 1170 | 1610 | 12.6 | 12.8 |
| | | 4-3-4 | 1270 | 1520 | 6.8 | 8.8 |
| | | 4-12-4 | 1290 | 1500 | 5.6 | 8.3 |
| | | 5-3-4 | 1290 | 1390 | 3.2 | 6.3 |
| | | 5-12-4 | 1270 | 1470 | 5.3 | 8.3 |
| | 650 | 1-5-3 | 1250 | 1640 | 10.5 | 10.9 |
| | | 3-4-3 | 1240 | 1650 | 9.5 | 8.9 |
| | | 3-5-3 | 1220 | 1600 | 10.5 | 9.9 |
| | | 4-4-3 | 1310 | 1540 | 4.2 | 4.2 |
| | | 4-5-3 | 1280 | 1650 | 7.4 | 7.4 |
| | | 5-4-3 | 1300 | 1610 | 6.3 | 7.3 |
| | | 5-5-3 | 1310 | 1470 | 2.1 | 4.2 |
| 1230° C/4 hr/RAC<br>+870° C/16 hr/RAC<br>+760° C/16 hr/AC | RT | 2-2-4 | 1280 | 1630 | 10.0 | 12.4 |
| | 650 | 2-4-3 | 1320 | 1750 | 7.9 | 7.3 |
| | | 2-5-3 | 1340 | 1760 | 9.5 | 9.3 |
| 1230° C/2 hr/RAC<br>+870° C/16 hr/RAC<br>+760° C/16 hr/AC<br>+870° C/1500 hr/AC | RT | 1-4 | 940 | 1260 | 8.4 | 10.2 |
| | | 1-5 | 960 | 1110 | 5.3 | 7.2 |
| | | 2-5 | 1010 | 1320 | 8.4 | 9.2 |
| | | 3-4 | 890 | 1120 | 5.3 | 6.7 |
| | | 3-5 | 900 | 1190 | 6.8 | 7.7 |
| | | 4-4 | 970 | 1130 | 3.2 | 5.2 |
| | | 4-5 | 970 | 1060 | 2.1 | 4.2 |
| | | 5-4 | 970 | 1020 | 2.1 | 3.1 |
| | | 5-5 | 970 | 1010 | 1.1 | 3.1 |
| | | [b]IIB-11 | 1050 | 1120 | 2.5 | 4.2 |

[a] RAC = rapidly air cooled; AC = air cooled.
[b] Cast and wrought NASA IIB-11 (ref. 1).

## TABLE IV. - AVERAGE ROOM-TEMPERATURE TENSILE PROPERTIES OF ALLOYS AFTER EXPOSURE FOR 1500 HOURS AT 870° C FOLLOWING PREFERRED HEAT TREATMENT

| Alloy | 0.2-Percent yield stress, MN/$m^2$ | Ultimate tensile strength, MN/$m^2$ | Elongation, percent | Reduction of area, percent |
|---|---|---|---|---|
| 1 | 950 | 1190 | 6.9 | 8.7 |
| 2 | 1010 | 1320 | 8.4 | 9.2 |
| 3 | 900 | 1160 | 6.1 | 7.2 |
| 4 | 970 | 1100 | 2.7 | 4.7 |
| 5 | 970 | 1020 | 1.6 | 3.1 |
| [a]IIB-11 | 1050 | 1120 | 2.5 | 4.2 |

[a] Cast and wrought NASA IIB-11 (ref. 1).

## TABLE V. - PERCENTAGE REDUCTION IN AVERAGE ROOM-TEMPERATURE TENSILE PROPERTIES OF ALLOYS AFTER EXPOSURE FOR 1500 HOURS AT 870° C FOLLOWING PREFERRED HEAT TREATMENT

| Alloy | 0.2-Percent yield stress | Ultimate tensile strength | Elongation | Reduction of area |
|---|---|---|---|---|
| | Reduction, percent | | | |
| 1 | 21 | 21 | 33 | 28 |
| 2 | 19 | 17 | 18 | 19 |
| 3 | 24 | 27 | 49 | 41 |
| 4 | 21 | 25 | 53 | 48 |
| 5 | 20 | 30 | 77 | 68 |
| [a]IIB-11 | 12 | 24 | 74 | 69 |

[a] Cast and wrought NASA IIB-11 (ref. 1).

## TABLE VI. - PHASES IN ALLOYS GIVEN PREFERRED HEAT TREATMENT AS DETERMINED BY ELECTROLYTIC EXTRACTION METHOD OF REFERENCE 7

| Phase | Alloy | | | | |
|---|---|---|---|---|---|
| | 1 | 2 | 3 | 4 | 5 |
| Hf concentration, wt % | <0.01 | 2.5 | <0.1 | 2.6 | 1.4 |
| C concentration, wt % | 0.088 | 0.084 | 0.205 | 0.190 | 0.142 |
| Amount of carbides, wt % | 2.9 | 2.5 | 4.2 | 4.2 | 3.2 |
| Types of carbides present: | | | | | |
| MC lattice parameter, $10^{-10}$ m | 4.314 | 4.540 | 4.405 | 4.420 | 4.483 |
| $M_6C$ lattice parameter, $10^{-10}$ m | 11.135 | 11.130 | 11.129 | 11.133 | 11.142 |
| Amount of gamma prime, wt % | 57 | 57 | 51 | 51 | 55 |
| Gamma-prime lattice parameter, | 3.595 | 3.590 | 3.588 | 3.589 | 3.591 |

## TABLE VII. - COMPOSITION OF METALLIC ELEMENTS IN CARBIDES EXTRACTED FROM ALLOYS GIVEN PREFERRED HEAT TREATMENT

| Element | Alloy 1 | | Alloy 2 | | Alloy 3 | | Alloy 4 | | Alloy 5 | |
|---|---|---|---|---|---|---|---|---|---|---|
| | wt % | at. % | wt % | at. % | wt % | at. % | wt % | at. % | wt % | at. % |
| Al | 0.4 | 1.5 | 0.6 | 2.4 | 0.3 | 1.2 | 0.2 | 1.0 | 0.3 | 1.3 |
| Ti | 1.6 | 3.4 | 1.0 | 2.3 | 3.8 | 8.7 | 1.8 | 4.9 | 1.7 | 4.0 |
| Ta | 19.2 | 10.9 | 25.6 | 15.5 | 36.1 | 21.8 | 32.6 | 23.5 | 27.5 | 17.2 |
| Hf | .5 | .3 | 19.9 | 12.2 | .5 | .3 | 20.7 | 15.2 | 8.9 | 5.6 |
| W | 38.8 | 21.7 | 21.3 | 12.7 | 29.3 | 17.4 | 22.2 | 15.8 | 31.3 | 19.3 |
| Mo | 12.1 | 13.0 | 6.0 | 6.8 | 9.0 | 10.3 | 6.5 | 3.6 | 9.3 | 11.0 |
| V | 1.1 | 2.2 | .7 | 1.5 | .9 | 1.9 | .6 | 1.5 | .8 | 1.8 |
| Zr | 1.4 | 1.6 | 2.1 | 2.5 | 1.1 | 1.3 | 1.4 | 2.0 | 1.1 | 1.4 |
| Cr | 8.5 | 16.8 | 6.3 | 13.3 | 6.5 | 13.7 | 4.8 | 12.1 | 6.6 | 14.4 |
| Co | 5.2 | 9.1 | 3.6 | 6.7 | 3.8 | 7.1 | 2.7 | 6.0 | 4.0 | 7.7 |
| Fe | .2 | .4 | .2 | .4 | .2 | .4 | .2 | .5 | .2 | .4 |
| Ni | 11.0 | 19.2 | 12.7 | 23.7 | 8.5 | 15.8 | 6.3 | 14.0 | 8.3 | 16.0 |

## TABLE VIII. - COMPOSITION OF GAMMA-PRIME PHASE IN ALLOYS GIVEN PREFERRED HEAT TREATMENT

| Element | Alloy 1 | | Alloy 2 | | Alloy 3 | | Alloy 4 | | Alloy 5 | |
|---|---|---|---|---|---|---|---|---|---|---|
| | wt % | at. % | wt % | at. % | wt % | at. % | wt % | at. % | wt % | at. % |
| Al | 6.3 | 14.2 | 6.1 | 13.7 | 6.6 | 14.7 | 6.5 | 14.5 | 6.1 | 13.7 |
| Ti | 1.1 | 1.4 | 1.1 | 1.4 | 1.0 | 1.3 | 1.2 | 1.5 | 1.1 | 1.4 |
| Ta | 10.2 | 3.4 | 9.2 | 3.1 | 7.9 | 2.6 | 6.4 | 2.1 | 7.7 | 2.6 |
| Hf | <.1 | <.1 | 2.3 | .8 | <.1 | <.1 | 1.7 | .6 | 1.3 | .4 |
| W | 6.5 | 2.2 | 4.3 | 1.4 | 7.4 | 2.4 | 6.8 | 2.2 | 6.6 | 2.2 |
| Mo | 1.1 | .7 | .6 | .4 | 1.1 | .7 | 1.2 | .8 | 1.3 | .8 |
| V | .7 | .8 | .7 | .8 | .6 | .7 | .7 | .8 | .7 | .8 |
| Zr | .1 | .07 | .1 | .07 | .1 | .07 | .1 | .07 | .1 | .07 |
| Cr | 2.3 | 2.7 | 1.9 | 2.2 | 2.3 | 2.7 | 2.4 | 2.8 | 2.3 | 2.7 |
| Co | 5.1 | 5.3 | 5.1 | 5.2 | 5.0 | 5.1 | 5.2 | 5.3 | 5.1 | 5.3 |
| Fe | .2 | .2 | .2 | .2 | .2 | .2 | .2 | .2 | .3 | .3 |
| Ni | 66.4 | 69.0 | 68.4 | 70.7 | 67.8 | 69.5 | 67.6 | 69.2 | 67.4 | 69.7 |

## TABLE IX. - COMPOSITION OF GAMMA PHASE IN ALLOYS GIVEN PREFERRED HEAT TREATMENT

| Element | Alloy 1 | | Alloy 2 | | Alloy 3 | | Alloy 4 | | Alloy 5 | |
|---|---|---|---|---|---|---|---|---|---|---|
| | wt % | at. % | wt % | at. % | wt % | at. % | wt % | at. % | wt % | at. % |
| Al | 2.5 | 5.5 | 2.9 | 6.7 | 2.8 | 6.2 | 3.2 | 7.2 | 3.2 | 7.1 |
| Ti | ---- | ---- | ---- | ---- | ---- | ---- | ---- | ---- | ---- | ---- |
| Ta | 1.3 | .4 | 1.9 | .7 | 3.0 | 1.0 | 4.7 | 1.6 | 4.1 | 1.4 |
| Hf | ---- | ---- | 1.7 | .6 | ---- | ---- | 2.0 | .7 | 1.0 | .3 |
| W | 6.6 | 2.1 | 10.7 | 3.6 | 5.5 | 1.8 | 6.6 | 2.2 | 6.4 | 2.1 |
| Mo | 2.1 | 1.3 | 3.3 | 2.1 | 1.9 | 1.2 | 2.0 | 1.3 | 1.9 | 1.2 |
| V | .5 | .6 | .4 | .5 | .5 | .6 | .4 | .5 | .5 | .6 |
| Zr | ---- | ---- | ---- | ---- | ---- | ---- | ---- | ---- | ---- | ---- |
| Cr | 18.2 | 20.7 | 18.1 | 21.6 | 16.8 | 19.2 | 16.2 | 19.0 | 17.8 | 20.6 |
| Co | 16.2 | 16.3 | 15.1 | 15.9 | 14.9 | 15.0 | 14.4 | 14.9 | 15.0 | 15.3 |
| Fe | .5 | .5 | .4 | .4 | .5 | .5 | .5 | .6 | .4 | .4 |
| Ni | 52.1 | 52.5 | 45.5 | 48.0 | 54.1 | 54.6 | 50.0 | 52.0 | 49.7 | 50.9 |

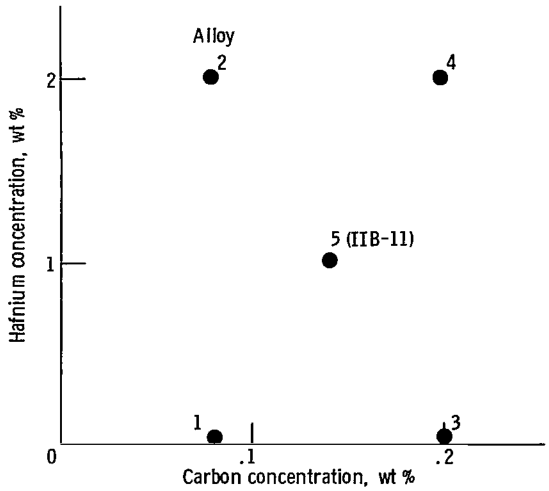

Figure 1. - Aim carbon and hafnium concentrations in alloys. (The number designations of the alloys are shown beside the points denoting the concentration.)

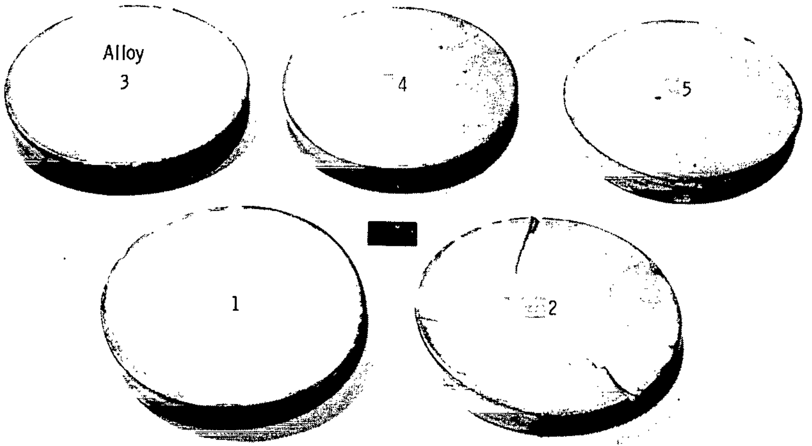

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Figure 2. - Appearance of cross-rolled disks identified by alloy number. (Torch marks on front edges of the disks indicate the direction of the last rolling pass.)

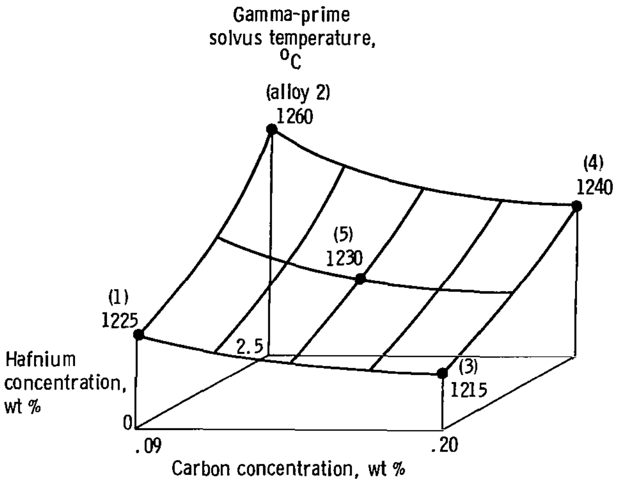


Figure 3. - Gamma-prime solvus temperatures of alloys as a function of carbon and hafnium concentrations.

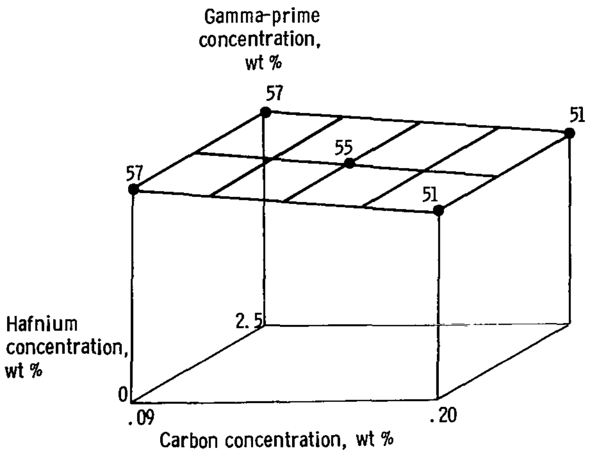
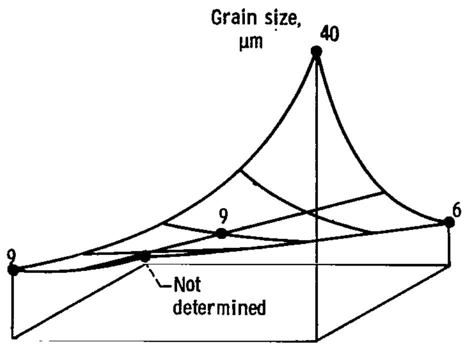
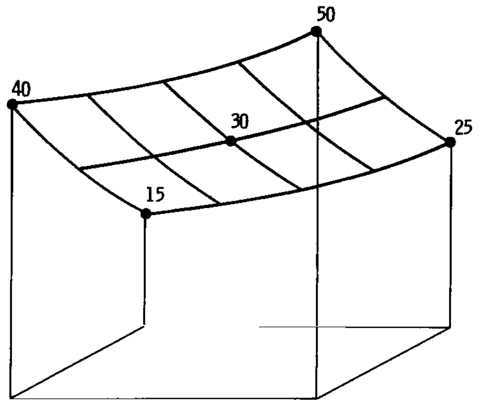


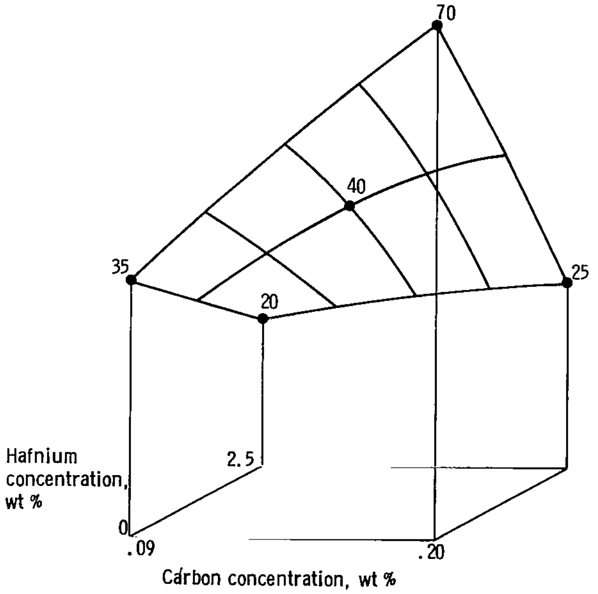
Figure 4. - Gamma-prime concentration in alloys as a function of carbon and hafnium concentrations.



(a) 1220° C/4 hr.



(b) 1230° C/2 hr.



(c) 1245° C/2 hr.

Figure 5. - Average grain sizes of alloys after heat treatments involving various solution treatments as function of their carbon and hafnium concentrations.

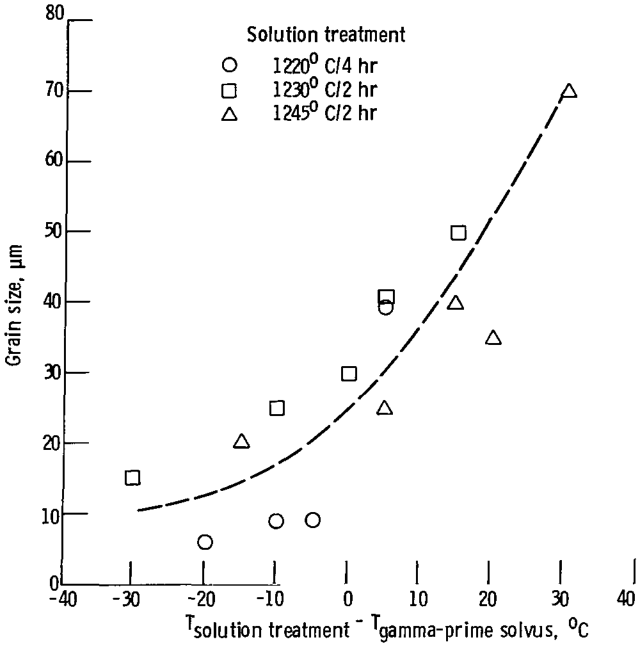
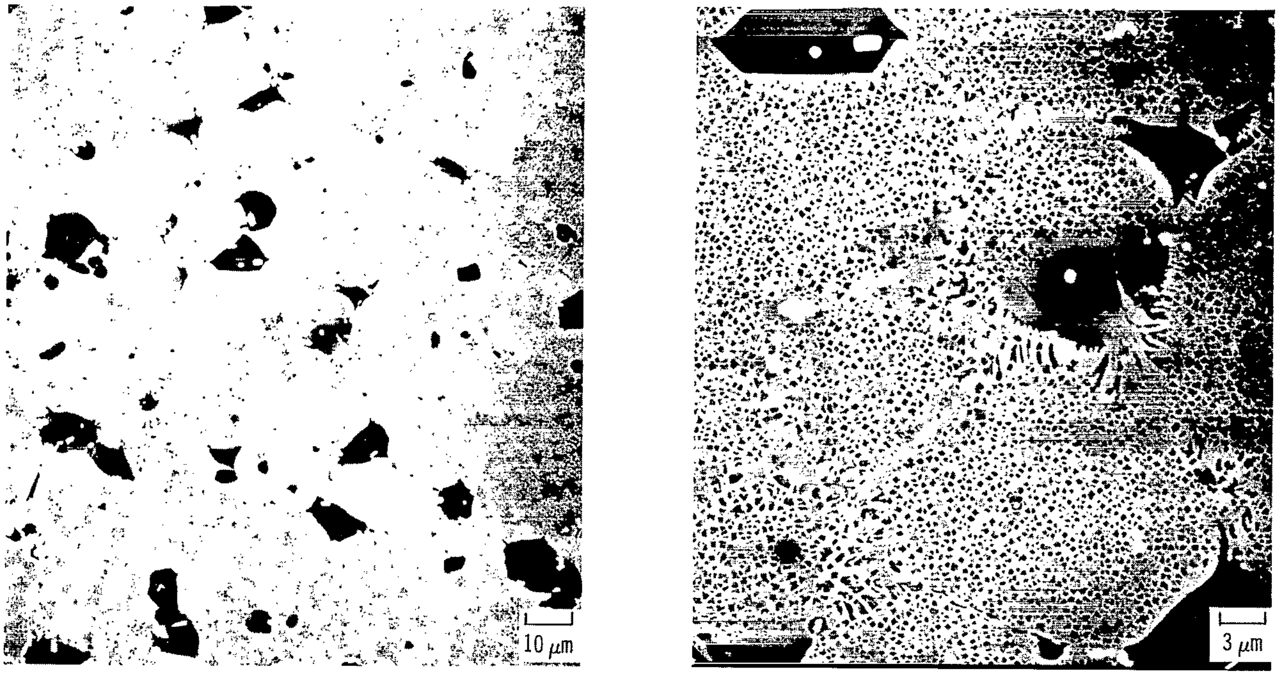
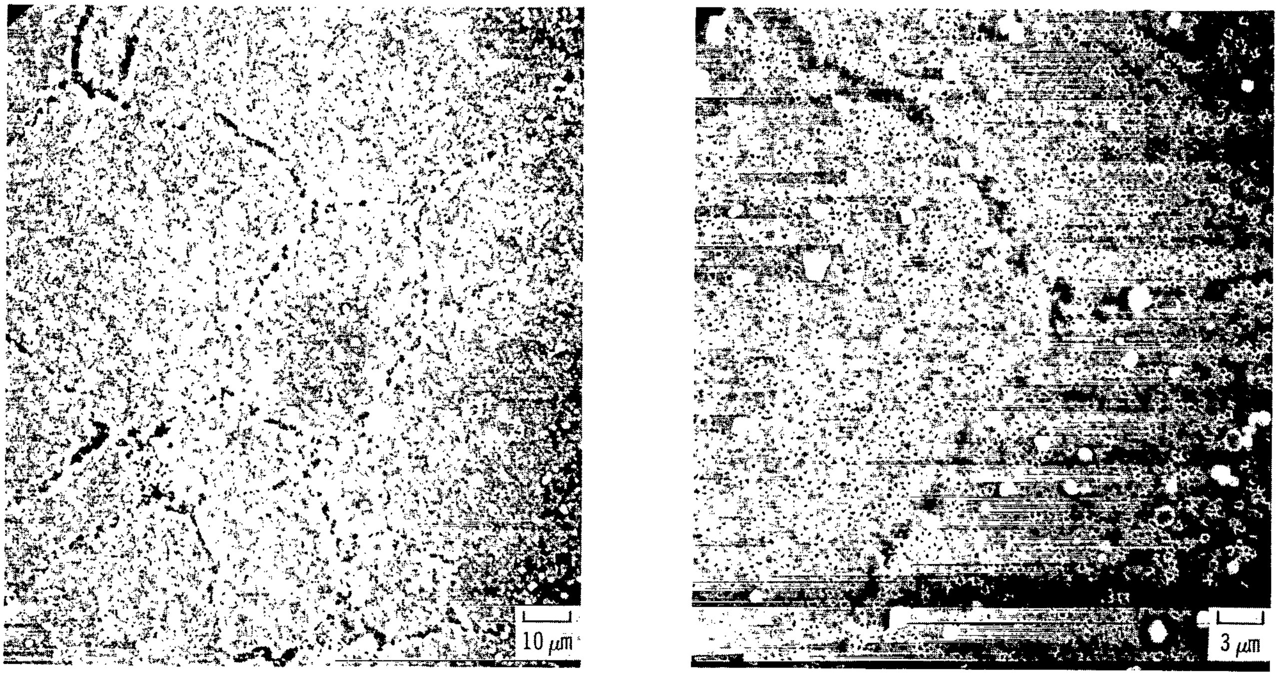


Figure 6. - Average grain sizes of alloys after heat treatments involving various solution treatments as function of the difference between the solution treatment temperatures and the alloys' gamma-prime solvus temperatures.



(a) Alloy 2.



(b) Alloy 5.

Figure 7. - Scanning electron micrographs of alloys given preferred heat treatment.

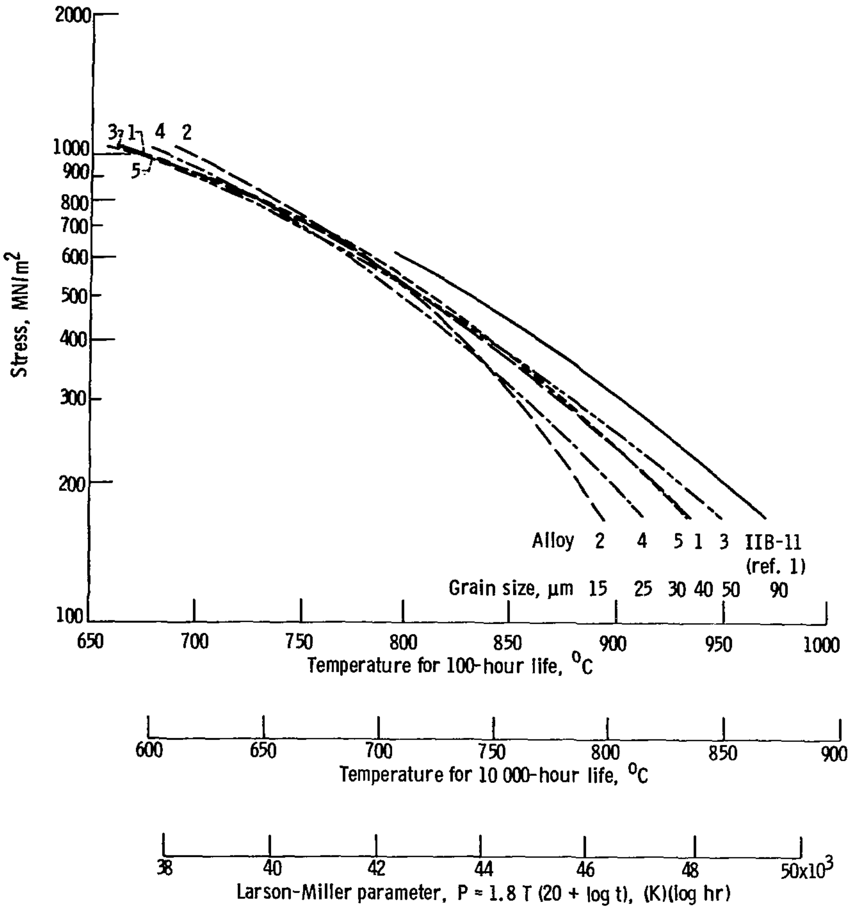
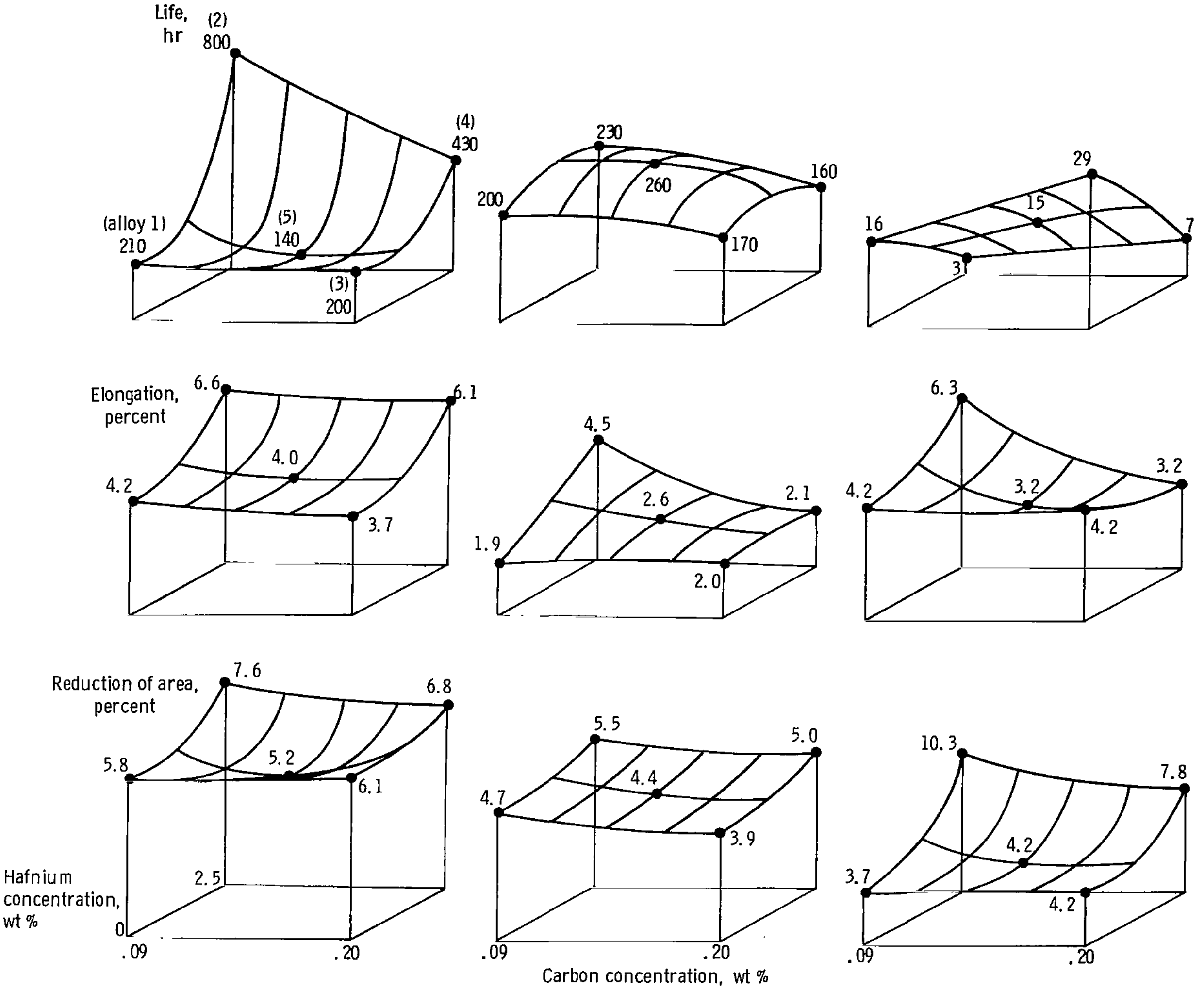


Figure 8. - Average stress-rupture data for alloys given preferred heat treatment plotted as stress against Larson-Miller parameter.



(a) Temperature, 650° C; stress, 1040 MN/m$^2$.

(b) Temperature, 760° C; stress, 620 MN/m$^2$.

(c) Temperature, 980° C; stress, 170 MN/m$^2$.

Figure 9. - Average stress-rupture properties of alloys given preferred heat treatment as function of their carbon and hafnium concentrations.

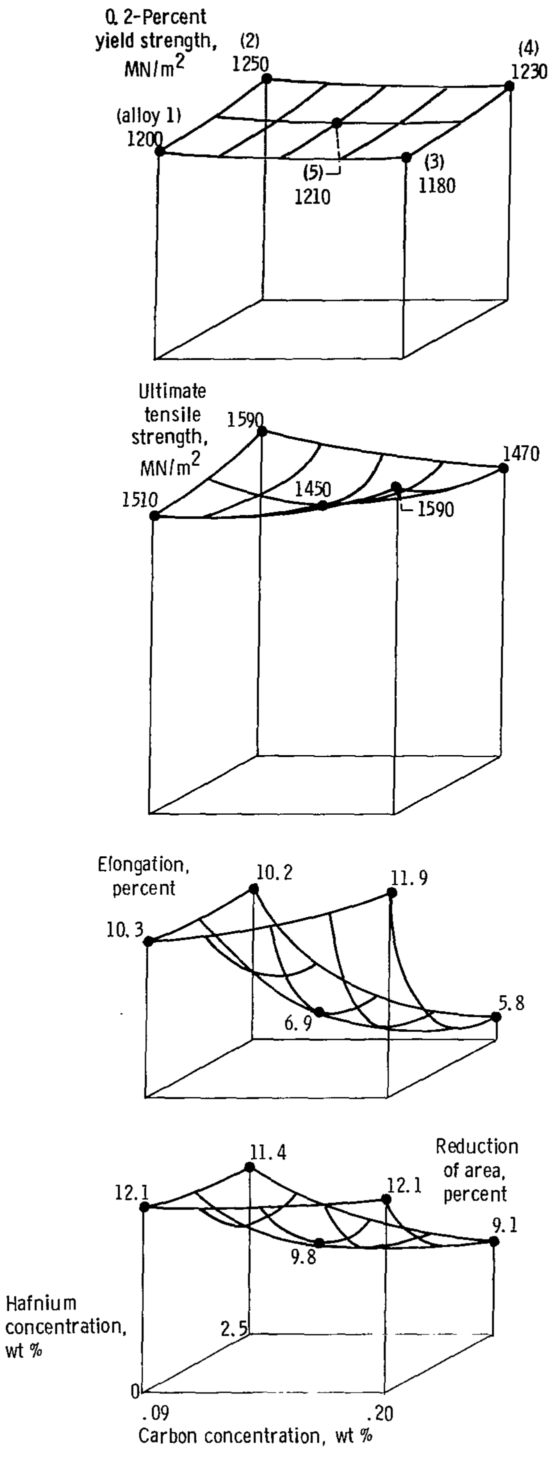


Figure 10. - Average room-temperature properties of alloys given preferred heat treatment as function of their carbon and hafnium concentrations.

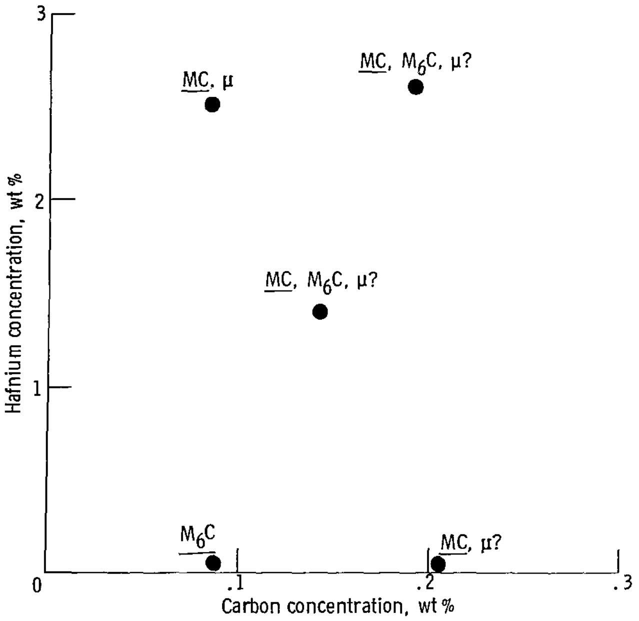
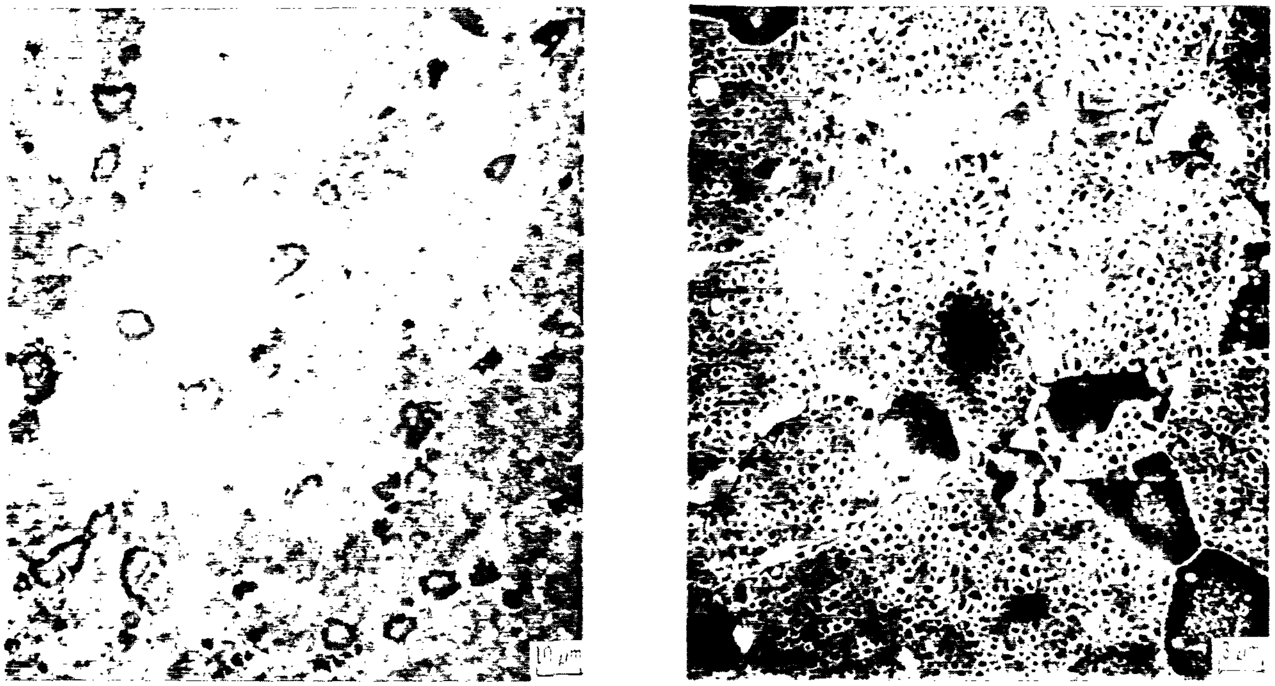
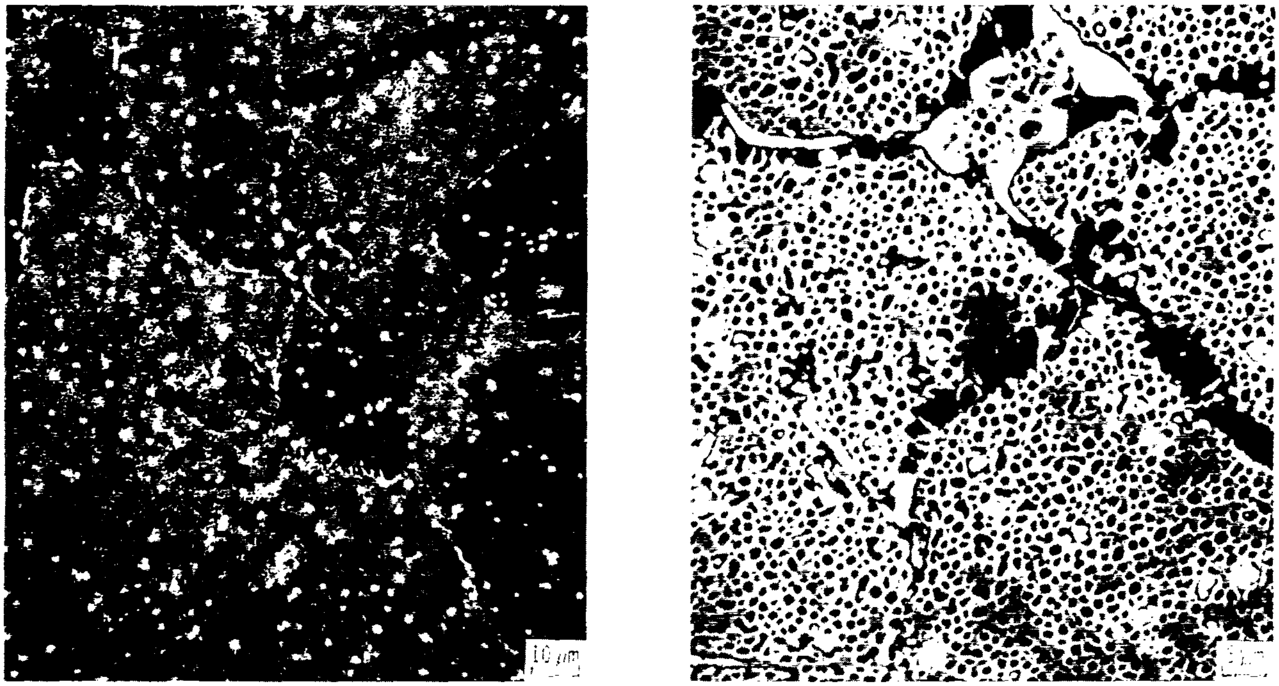


Figure 11. - Minor phases in alloys given preferred heat treatment and exposed for 1500 hours at 870° C, as function of carbon and hafnium concentrations in the alloys. (The most abundant phases are underlined.)



(a) Alloy 2.



(b) Alloy 5.

Figure 12. - Scanning electron micrographs of alloys exposed for 1500 hours at 870° C after preferred heat treatment.

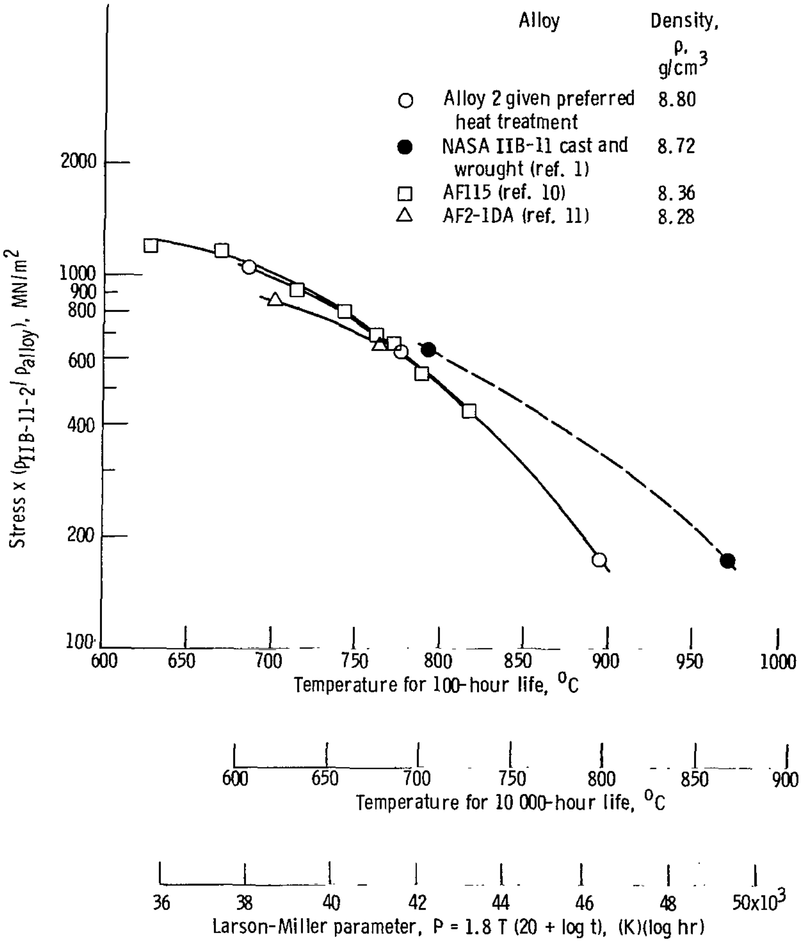


Figure 13. - Comparison of stress-rupture strengths of four candidate materials for a 750° C turbine disk.

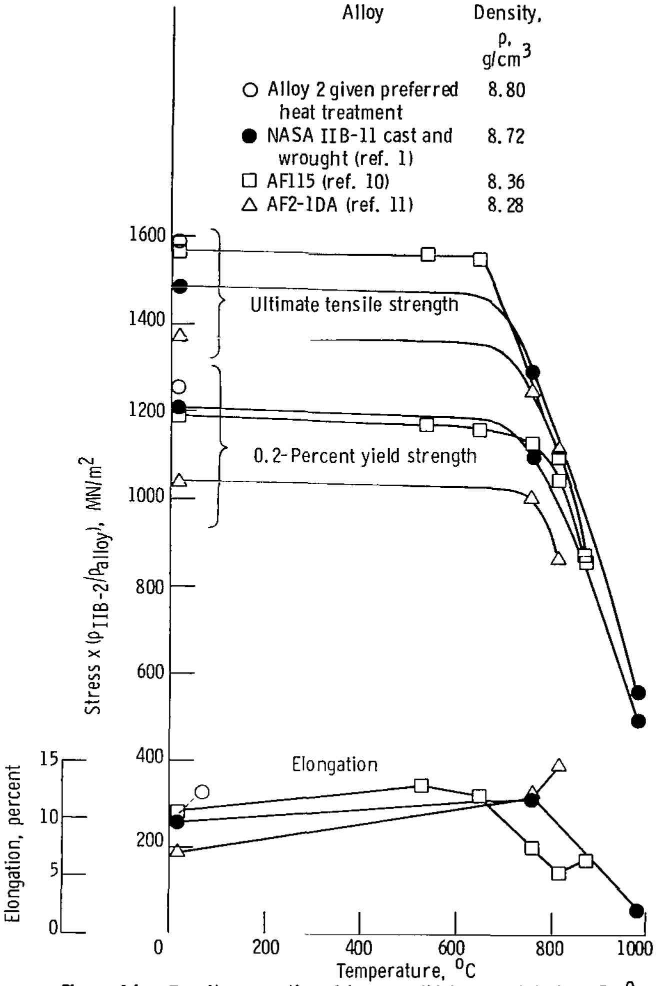
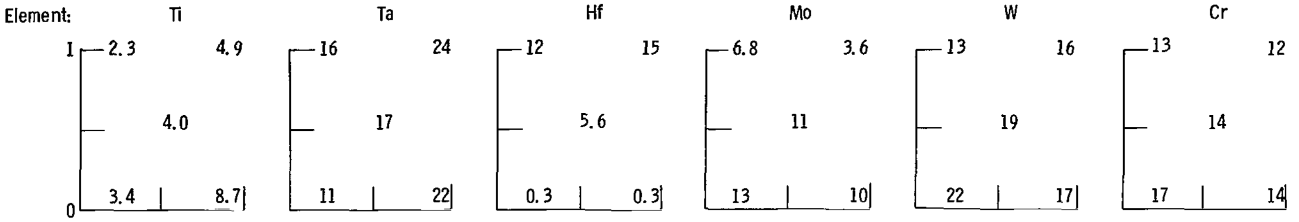
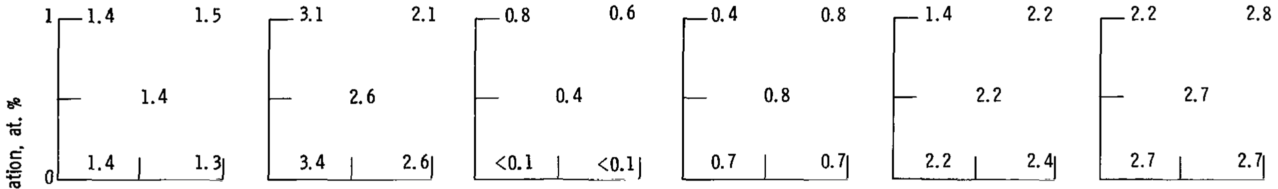


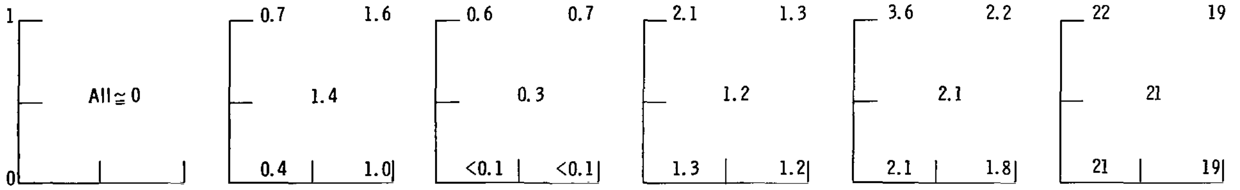
Figure 14. - Tensile properties of four candidate materials for a 750° C turbine disk.



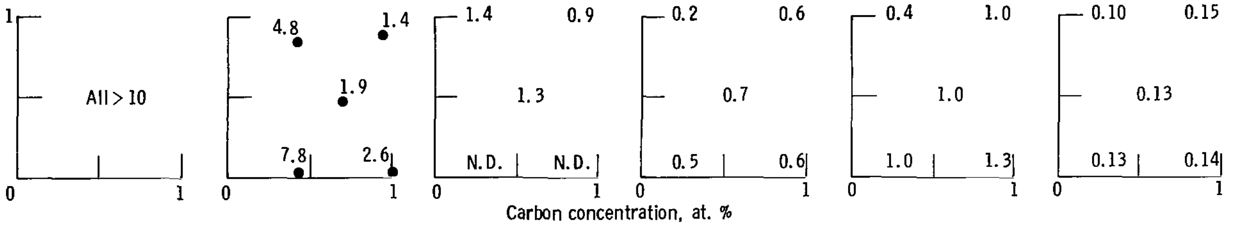
(a) Concentration in carbides, at. % of metallic elements only.



(b) Concentration in gamma prime, at. %.



(c) Concentration in gamma, at. %.



(d) Ratio of concentration in gamma prime to concentration in gamma.

Figure 15. - Concentrations of selected elements in alloy phases shown as function of carbon and hafnium concentrations of the alloys. (N.D. denotes not determined.)

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