

ENGINEERING RESEARCH INSTITUTE  
THE UNIVERSITY OF MICHIGAN  
ANN ARBOR

Progress Report No. 1

IRON-BASE HIGH-TEMPERATURE ALLOYS

August 15 to November 15, 1957

(This project is a continuation of Project 2313 and therefore this report is the eleventh quarterly report on this alloy development program.)

R. W. Kraft  
D. L. Sponseller  
R. A. Flinn

Project 2698

CHRYSLER CORPORATION  
DETROIT, MICHIGAN

December 1957

## ABSTRACT

A significant improvement in the combination of high- and low-temperature properties obtainable in UM 233 has been achieved recently by a short heat treatment at 2150°F. It was found that a one-hour soaking period at this temperature increased the room-temperature ductility of an 18% Ni, 18% Cr, 0.3% C, 1.25% B, 5.0% Mo alloy without decreasing the rupture strength below 30,000 psi (1500°F). Appreciable plastic deformation compared with untreated material (3% elongation) occurred at room temperature before failure (see Fig. 4).

It has also been found that a more extended heat treatment raises the rupture strength of Fe-Ni-Cr-B alloys. The same extended treatment lowers the rupture strength of alloys containing in addition the elements C, Mo, and W. It may be possible, therefore, to achieve an even better combination of properties by replacing the above hardening elements with a heat treatment.

Because molybdenum impairs the oxidation resistance while improving high-temperature properties, some work has been done with substituting tungsten for molybdenum. In the as-cast condition, the following analysis has given a 100-hr rupture strength at 1500°F in excess of 30,000 psi with an appreciable room-temperature ductility (1.2% elongation)—18% Ni, 18% Cr, 0.2% C, 1.25% B, 3% Mo, 4% W.

A comparison of analyzed boron contents of samples from the same heats performed at different laboratories indicates that the Chrysler Corporation ion-exchange method produces consistent, meaningful results. In the past the boron-analysis problem has been a source of considerable trouble.

## OBJECTIVE

The objective of this research project is to develop an analysis specification, melting procedure, and heat treatment which will produce a 100-hr rupture strength at 1500°F of 30,000 psi combined with room-temperature tensile strength of 70,000 psi and as much ductility as possible (10% minimum elongation preferred). The material so developed should be low in cost and strategic alloy content and permit manufacture with a minimum of special techniques.

## INTRODUCTION

Officially, this is the first quarterly report on Engineering Research Institute Project 2698, but since the actual work is a continuation of that initiated under Project 2313 three years ago, this report is actually the eleventh quarterly report. It outlines the progress which has been made in the last three months in developing an iron-base high-temperature alloy.

The report is divided into three main categories for discussion purposes. The reader should realize, however, that all three phases are closely related and are being carried on simultaneously. The report is concluded with a brief resumé of plans for the immediate future.

Mechanical property data acquired within the last three months are given in Appendix Table I at the end of this report.

## DISCUSSION OF RESULTS

### I. ALLOY DEVELOPMENT—AS-CAST PROPERTIES OF UM 233\*

Until quite recently, the objective of a large portion of the experimental work has been to establish chemistry specifications which will yield a minimum of 30,000-psi 100-hr rupture strength at 1500°F with as much room-temperature ductility as possible—in the as-cast condition. This section of the report is limited to a discussion of recent results on as-cast materials. Heat-treatment studies are included in the next section.

A. Molybdenum.—All stress-rupture tests have now been completed on two series of heats (one with 0.3% C, 1.25% B, the other with 0.5% C, 1.25% B) designed to delineate the optimum molybdenum content of UM 233 containing 18% Ni and 18% Cr. In addition actual chemical analyses have been determined. All these new data confirm the tentative conclusions reached in the previous report, namely:

1. A molybdenum content of 4.5% minimum is needed to achieve a 30,000-psi 100-hr rupture strength at 1500°F.
2. The higher carbon—higher boron alloys are not as ductile as the lower carbon—lower boron alloys.

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\*UM 233 is an iron-base alloy containing Ni and Cr plus C, B, Mo, and W as the principal strengthening elements.

In short, Fig. 2 of the previous report is essentially unaltered by the new data.

B. Tungsten.—Previous progress reports have mentioned that tungsten is one element which apparently has no deleterious effects on these alloys (with the exception of its cost and strategic value) in amounts up to 1% W. In fact, it has increased the rupture strength, especially at the more desirable low carbon levels. With these thoughts in mind additional heats have been prepared.

The data now available are summarized in Fig. 1. Caution is needed in accepting the improvement indicated by the dashed line on the graph because:

1. The rupture strengths for the higher tungsten heats are based upon only one or two tests which have been completed.
2. Carbon is a critical variable (steep slope to the line), and since the carbon analyses are not yet available on some of the heats, the plotted points may be displaced.
3. Actual tungsten analyses on some heats are still missing. On one heat in particular, R391, the tungsten analysis must be rechecked to establish whether the carbon series at the 2% W level does exhibit such a marked increase in strength with carbon—viz.:

<u>Heat</u>	<u>% C</u>	<u>100-hr R.S.</u>
R391	.14	24,700
R401	.20 aim	~ 27,500
R403	.30 aim	~ 30,000

C. Thermal-Shock Test Results.—Several wedge-shaped thermal-shock test samples from three different heats of UM 233 were prepared and submitted to Chrysler Corporation for evaluation of results. Basically the test consists of very rapidly heating and cooling a sample and thus inducing severe thermal gradients and thermal stresses. By noting the number of cycling tests required to cause a crack or complete failure, an empirical index of the relative "thermal-shock resistance" of different materials can be made. The data in Table I are reproduced from the Chrysler report together with representative values for Haynes Stellite 31 (Co-base high-temperature alloy) and ceramic materials. The data are repeated here for completeness.

The conclusions which may be drawn are as follows.

1. Compared to ceramic materials, UM 233 would most definitely not be classified as a brittle material or one possessing inferior thermal-shock resistance.
2. Compared to HS 31, the iron-base alloys do not exhibit quite as much thermal-shock resistance as measured by this test. Whether the iron-base UM 233 alloys will perform satisfactorily in turbine applications can only be determined by actual turbine tests.

3. Among the three UM 233 alloys studied, there is probably no statistically significant difference. Heat R372 may be slightly better than the other two but the difference is small. Since the alloys were chosen to evaluate different carbon, boron, and tungsten levels, it is probably safe to say that the thermal-shock resistance will not be a limiting feature in this alloy development program.
4. All the UM 233 alloys exhibited somewhat more oxidation during the test than is usually found in HS 31 alloy.

The last conclusion was the incentive for the next section of this report.

D. Molybdenum—Tungsten.—It is a well-known fact that molybdenum has poor oxidation resistance at elevated temperatures. The application of this fact to the present alloy development program was demonstrated several years ago when it was observed that the molybdenum content of UM 233 must be kept below 5%, even though higher rupture strengths could be obtained at 15% Mo. In fact, a large part of the work within the past year has been devoted to finding a substitute for the extra 10% of Mo which would cause an unacceptable increase in the rate of oxidation.

The present series of heats was made with the objective, therefore, of decreasing the molybdenum content below the 4.5-5.0% level described in a previous section of this report. The data presented there indicated that if the molybdenum content is reduced to 3%, the 100-hr rupture strength would fall to about 27,000 psi in tungsten-free alloys. Consequently it was decided to add tungsten in view of the promising (though not conclusive) trend indicated in Section I-B and Fig. 1. Pertinent data now available are shown in Table II.

Many of the critical tests are still in progress and most of the low Mo, high W rupture strengths reported in the table are based on only one tensile bar. The following encouraging trends are noted, however.

1. In two cases, R408 and R409, an estimated 100-hr rupture strength of at least 30,000 psi was obtained with only 3% Mo. Less tungsten is needed at the higher carbon content, as would be anticipated.

	<u>R408</u>	<u>R409</u>
% C	.2	.3
% W	4.0	2.0
Est. 100-hr R.S.	31,000 psi	30,000 psi

2. If the other stress-rupture tests now in progress on R408 confirm the tentative conclusion enumerated above, this particular analysis would be one of the most desirable analyses (with the one exception of its high tungsten content) produced to date.

R408 Analysis	{ 18% Ni, 18% Cr, 1.25% B 0.2% C - low carbon for better weldability 3.0% Mo - low molybdenum for superior oxidation resistance 4.0% W - high tungsten an economic disadvantage
1500°F Rupture Strength	31,000 psi estimated 100-hr life
Room-Temp. Properties	81,400 psi tensile, 1.2% el., 4.5% red. area

The room-temperature stress-strain curve of this material is shown in Fig. 2. It is not an extremely ductile material, but the curve does show that some plastic deformation occurs before failure.

- By comparing vertical columns in Table II it is noted that tungsten apparently has a more pronounced beneficial effect at the lower carbon levels. These data confirm the results mentioned in a previous report and are based on a different series of heats.

E. Nickel.—Two alloys (R410, R411) have been made at the 14.0% Ni level to determine if a saving of 4.0% of Ni can be realized. Not even tentative conclusions can be drawn at this time because the stress-rupture tests are still in progress. It will be recalled that it has been previously shown that the high-temperature properties fall off slightly if the nickel content is cut to 10% (from 18% Ni).

F. Chemical Analysis.—Chemical analysis for all alloying elements other than boron is a fairly routine procedure and need not be discussed here. There has been some trouble with boron analyses in the past, however, so it is well to review the status of this problem. The pertinent data, including all new data applicable to this problem which have been acquired within the past three months, are included in Table III.

The first part of the table gives the analyses reported by three different laboratories on samples from the same heats. It should be noted that the Pitkin laboratory method and the new Chrysler ion-exchange method agree quite well over the entire range of boron contents. Since these results are compatible with each other and with the known added boron, it can be said that the Chrysler chemists have developed a reproducible technique. The Colburn laboratories had considerable difficulty with the samples and in two cases (R311 and R308) reported a boron recovery well in excess of 100% recovery. These facts make the reliability of their determinations doubtful.

The second part of the table shows values reported by Chrysler for a series of UM 233 alloys. In view of the extremely difficult analysis problem and since the actual boron content of the heats might vary considerably between heats, these values are considered to be a further indication that a satisfactory method has been developed. Some idea of the probable error in reported boron con-

tent can be obtained by statistically analyzing the data for 1.25% B heats. The average value reported is about 1.12% and the standard deviation is .10% B.

Thus it can be stated that the probable actual boron content of the nominal 1.25% boron heats is approximately  $1.12 \pm .07\%$  B. This corresponds to a boron recovery of about 84 to 95%.

The sponsor may want to consider running several determinations on one or two heats to obtain an idea of the reproducibility of the analysis itself without having the added variable of heat-to-heat differences. This type of information should prove extremely useful for control purposes in the future and would also facilitate the alloy development work.

## II. HEAT-TREATING EXPERIMENTS

The immediate objective of these experiments is to develop a heat treatment which will increase the room-temperature properties, especially the ductility, of UM 233 without decreasing the 1500°F rupture properties. From a long-range standpoint, it is of course desirable to understand what is happening and why, so that further improvements may be made.

Because this is a new phase of the work, a brief outline of the experimental procedure should be given.

1. Mold making, dipcoating, investing, dewaxing, firing, melting (under argon), pouring, cleaning, and radiographic and visual inspection of all test bars which are to be tensile- or rupture-tested follow exactly the same standard procedure which has been rigidly adhered to in this research program for the last three years.

2. The good castings (.250-in.-diam tensile bars) which are to be heat-treated are loaded into a muffle built into a gas-fired furnace. A platinum-platinum-10% rhodium thermocouple connected to a Brown recorder is placed next to the bars after which the muffle is sealed. The seal is so arranged that argon gas is passed through the muffle during heat treatment to prevent oxidation of the specimens at the elevated temperatures employed. The furnace controls are so regulated that the control thermocouple next to the specimens records the desired value.

3. After the specified heat treatment has been completed, the seal is broken, and the tensile bars are removed with a wire and allowed to cool in still air.

4. For all the alloy development program to date, normally two and usually three or four stress-rupture bars are tested at different loads to establish more accurately the 10- and 100-hr rupture strength. But for these experiments a short cut has been adopted for the preliminary work at least. Only one bar is tested for any given set of conditions and the rupture life deter-

mined for a given load. The actual stress-rupture and tensile testing is done, as usual, according to applicable ASTM standards.

Two "standard" analyses of UM 233 have been chosen for the preliminary work. The first is an alloy which contains the minimum amount of strengthening elements to give a 30,000-psi 100-hr life at 1500°F. The second is one with more alloy and therefore stronger. It was chosen because it was anticipated that a heat treatment which increased the room-temperature ductility would probably also decrease the elevated temperature properties. These two analyses, which will be called UM 233A and UM 233B for brevity, and their as-cast properties are described on the following page. From these data, it was decided to proof-test UM 233A heats at 30,000 psi and UM 233B heats at 33,000 psi. In the as-cast condition this should give about a 100-hr life at 1500°F in each case, with 13.0 and 8.0% total elongation, respectively.

A. Effect of Solution Heat-Treating on Properties.—In report 2313-9-P it was shown that a 24-hr solution treatment at 2150°F caused a decrease in the rupture time (at 30,000 psi) and rupture elongation on one sample of UM 233A. It was decided, therefore, to investigate the effect of solution-treating time and temperature before proceeding with any aging experiments.

Stress-rupture bars (and microspecimens) of both alloys were heated at 2150°F for periods of 1, 5, and 24 hours with the results shown in Fig. 3. Microspecimens only were also heat-treated at 2000°F and 1800°F for the same times. These data will be discussed in the next section.

Interesting trends revealed by the plot of data in Fig. 3 are as follows:

1. For both alloys, an increasing time at the solution-treating temperature progressively decreased the rupture life. Whether this is due to a lack of hot ductility or not is not clear from the data. It should be noted that the rupture lives of two sets of duplicate data showed remarkable reproducibility.

2. Contrary to expectations, the maximum room-temperature ductility was obtained with the shortest heat treatment. The stress-strain curves of Fig. 4 illustrate this phenomenon clearly. The increase in ductility is more pronounced in the lower carbon, tungsten-free analysis, but the same trend holds also for the UM 233B alloys.

These curves also illustrate the fact, which has been mentioned before, that in the more brittle alloys, it is the lack of ductility which is responsible for low tensile strengths. The more ductile alloys usually have better strength and thus the room-temperature tensile strength is also an indication of ductility.

The results on the bars of UM 233A solution-treated one hour at 2150°F are the best combination of mechanical properties which have been produced to date,



Analysis	UM 233A				UM 233B
	18% Ni, 18% Cr, 0.3% C, 1.25% B, 5% Mo	18% Ni, 18% Cr, 0.5% C, 1.25% B, 5% Mo, 1% W			
Heat Nos.	R351	R364	R369	R383	R353
1500°F, 100-hr Rupture Strength, psi	29,800	~ 29,600	~ 29,000	~ 31,000	29,800
Life, hr, and (elong, %) at 30,000 psi	96(14.0)	92(12.0)			94(13.0)
at 33,000 psi					100(8.0) estimated
at 35,000 psi	24.1(11.0)		11.2(9.5)	46.9(12.0)	27.4(11.0)
R.T. Tensile Strength	94,200				62,300
Elongation %	1.5				< 1.0

as shown below.

Analysis, UM 233A	18% Ni, 18% Cr, 0.3% C, 1.25% B, 5% Mo.
Heat Treatment:	2150°F - 1 hr - air cool.
1500°F Properties:	30,300-psi estimated 100-hr rupture strength.
Room-Temp. Properties:	76,700-psi tensile strength with <u>3.0%</u> elong, <u>4.5%</u> red. area.

Duplicate tests are being obtained on another heat of this same analysis to confirm these data. In addition a pair of bars will also be solution-treated for 15 minutes.

B. Effect of Solution Heat-Treating a Microstructure.—As mentioned before, small samples from both analyses have been heat-treated at 1800°F, 2000°F, and 2150°F for periods of 1, 5, and 24 hours in argon. A portion of these structures is shown in Figs. 5 and 6 (Fig. 5: UM 233A - As-cast, 1 hr at 1800°F, 2000°F, 2150°F; Fig. 6: UM 233A - As-cast, 1, 5, and 24 hr at 2150°F). These photomicrographs illustrate the trends which have been noticed in all the samples, namely:

1. Fig. 5—Higher temperatures, at a constant heat-treating time, cause an increase in the degree of spheroidization of the as-cast gross precipitate. This effect is not pronounced until the temperature exceeds 2000°F (1-hr treatment).

2. Fig. 6—Increasing time, at a constant temperature, increases the degree of spheroidization. In a previous report, it was shown by means of the microbend apparatus that this spheroidization caused a local increase in the ductility. The actual tensile data on heat R405 (Fig. 4), quite surprisingly, showed that maximum tensile ductility was obtained before spheroidization had progressed to a high degree. At the present time no explanation for this anomaly is apparent. The problem is being studied from several angles. In all probability a fine precipitate shown in some pictures plays an important role.

3. At the present time there seems to be no direct relation between microstructure and rupture strength. It is true that the rupture strength decreases as the degree of spheroidization increases, but, as will be pointed out in Part III of this report, it is thought that the matrix is more influential than the precipitate in governing the rupture strength. Microhardness work now in progress may clarify the picture to some extent. The correlation of microstructure with plastic strain will be observed microscopically with the "Strain Viewer" and compared with HS 31.

### III. FUNDAMENTAL DEVELOPMENT WORK

A few of the experimentally verified facts concerning the role of boron in these alloys are as follows.

1. Boron in amounts up to 2.0% added causes an increase in the rupture strength of alloys containing only iron, 18% Ni, and 18% Cr.

2. Boron in amounts up to 2.0% added leads to the formation of gross boride precipitate phases, the amount of which is almost directly proportional to the amount of boron in the alloy. These borides are thought to be a complex mixture of  $Fe_2B$  and  $Cr_2B$  and perhaps other borides.

3a. Boron in amounts up to 2.0% added causes an increase in the hardness of the austenite.

3b. The austenite can be softened (by a 24-hr heat treatment at 2200°F) to a level characteristic of the boron-free alloys—regardless of the original boron content. During this heat treatment some spheroidization of the borides occurs.

These conclusions are the result of a series of experiments on high-purity vacuum-melted heats in which boron was the only strengthening agent used.

From the preceding, one cannot tell exactly whether it is the boride or the matrix which is predominant in governing the rupture strength since a linear correlation can be obtained between items 2 and 1 and also between items 3a and 1. It was decided, therefore, to rupture-test bars which had previously been solution-treated to the extent that their matrix hardness prior to test was in the completely softened condition. It was reasoned that the strength would decrease if the matrix hardness were the governing factor, or would remain the same if the amount of boride were the governing factor, since the amount of gross boride is essentially unchanged by the solution treatment.

The results of those experiments on which data are now available are summarized below.

	<u>Bars Tested in the</u> <u>As-Cast Condition</u>	<u>Bars Tested After a</u> <u>2150°F, 24-hr Solution Treatment</u>
0.1% B heats - Tests in progress		
0.5% B heats, 8500-psi load in 1500°F rupture test		
R344 - Life	17.0 hr	382 hr
Elong	25.5%	24.0%
R396 - Life	108.1 hr	445 hr
Elong	23.0%	18.0%
1.0% B heat, 12,000-psi load in 1500°F rupture test		
R395 - Life	17.8 hr	48.7 hr
Elong	32.0%	39.0%

Notice that in every case the rupture life (and therefore the rupture strength) increased as a result of the solution treatment, in contrast to the UM 233 alloys which exhibited a decrease in rupture time after solution-treating. These results lead to several interesting ramifications.

1. Microspecimens have been cut from the gage lengths of the 0.5% B samples which were solution-treated prior to test. A fine precipitate was noted in the structure and the matrix hardness had again increased. This then is a very strong indication that it is the matrix (as contrasted to the amount of gross boride present) which governs the rupture strength, and furthermore that some type of aging or strain-aging mechanism involving boron strongly influences the strength.

X-ray diffraction patterns of the residues of various samples confirm these results because different diffraction lines have been observed on "before and after rupture-testing" samples of tensile bars tested in the as-cast condition.

Obviously more work needs to be done to present a more coherent picture of the mechanism, but the essential idea that the rupture strength is governed by some type of aging phenomenon (apart from the gross boride content) is thought to be valid.

2. Theoretically speaking at least, if it is the boron in the matrix which governs the high-temperature properties (this is what dislocation theory predicts and what the preceding paragraphs show), it should be possible to obtain high rupture strengths at quite low boron contents by appropriate heat treatment. Practically speaking, it may take an extended period of time to develop the appropriate conditions which theory predicts.

3. The opposite behavior exhibited by Fe-Ni-Cr-B alloys and those containing in addition Mo and C (UM 233 analyses) is quite probably connected with the carbon content. Recent work at the University on other sponsored research has shown that the two elements, carbon and boron, interact with one another and that this interaction has a profound effect on the high-temperature properties of some metals. This other work, although not directly applicable to the present problem for a number of reasons, focuses attention on a possible cause for the observed difference in behavior between carbon-free and carbon-containing alloys in this work.

For the purposes of the practical alloy development program, a carbon-free or low-carbon alloy would be more desirable. Several argon-melted, carbon-free heats, containing molybdenum and tungsten, have been prepared. These heats will be used for solution heat-treating experiments in the hope that a heat treatment can be developed which will give high rupture strength and better room-temperature ductility. This alloy should also have better weldability because of a much lower carbon content.

## FUTURE WORK

The plans suggested throughout this report for the immediate future may be very briefly summarized as follows.

1. Repeat solution-treating series on another heat of UM 233A to confirm the encouraging results shown in Figs. 3 and 4.
2. Extend this series to shorter solution-treating times.
3. Develop data on other solution-treating temperatures.
4. Depending on the results of items 1, 2, and 3, experiment with solution-treating and preaging treatments.
5. Repeat item 1 (and 2 and 3 and 4 as seems necessary) for carbon-free alloys.
6. Endeavor to establish data which will clarify the role of boron in iron-nickel-chromium alloys.

Obviously, in addition, tentative conclusions based on limited data must be verified or rejected when all test results are in.

APPENDIX TABLE I  
 PROPERTIES OF HEATS POURED IN ALLOY PROGRAM  
 (Data acquired since September 1, 1957)

Heat No.	Nominal Composition	Remarks	Heat Treatment <sup>(4)</sup>	Test Temp, °F	Mechanical Properties				1500°F Summary Properties		
					Stress, psi	Life <sup>(1)</sup> , hr	Elong, %	Reduction Area, %	Hot Tensile Strength, psi	Rupture Strength, psi	
										10 hr	100 hr
R309	18% Ni, 18% Cr, 2.0% B	Vacuum-melted	As-cast	1500	16,000	39.1	13.0	30.0			
R344	18% Ni, 18% Cr, 0.5% B	Vacuum-melted	As-cast	1500	8,500	17.0	25.5	69.0			
			As-cast	1500	6,800	155	20.5	59.0			
			2150-24-AC <sup>(4)</sup>	1500	8,500	382	24.0	42.0			
R383	18% Ni, 18% Cr, 0.3% C, 1.25% B, 5.0% Mo	Repeat 369, 351	As-cast	1600	16,500	308	10.0	16.5	50,600 (1600°F)	31,000 <sup>(3)</sup> (1600°F)	20,200 <sup>(3)</sup> (1600°F)
R388	18% Ni, 18% Cr, 0.3% C, 1.25% B, 3.0% Mo	Low Mo UM 233	As-cast	1500	26,000	156	16.0	21.5	52,500	33,900 <sup>(3)</sup>	27,100 <sup>(3)</sup>
R389	18% Ni, 18% Cr, 0.3% C, 1.25% B, 7.0% Mo	High Mo UM 233	As-cast	1500	35,000	29.0	10.0	11.0	61,300	40,000	30,100
R390	18% Ni, 18% Cr, 0.2% C, 1.25% B, 5% Mo, 0.28% V	Vanadium heat	As-cast	1500	25,000	240	16.0	16.5	57,300	35,800	27,500 <sup>(3)</sup>
R395	18% Ni, 18% Cr, 1.0% B	Vacuum-melted	As-cast	1500	12,000	17.8	32.0	41.5			
			2150-24-AC	1500	12,000	48.7	39.0	50.0			
R396	18% Ni, 18% Cr, 0.5% B	Vacuum-melted	As-cast	1500	8,500	108.1	23.0	49.5			
			2150-24-AC	1500	8,500	445	18.0	50.5			
R400	18% Ni, 18% Cr, 0.5% C, 1.25% B, 5% Mo, 1% W		2150-5-AC	1500	33,000	30.3	5.0	8.0			~ 28,000
			2150-24-AC	1500	33,000	19.3	6.0	10.5 <sup>(2)</sup>			~ 27,000
R401	18% Ni, 18% Cr, 0.2% C, 1.25% B, 5% Mo, 2% W	.2 C, 2 W	As-cast	1500	58,000	STTT	7.0	8.0	58,000		~ 27,500
			As-cast	1500	30,000	47.6	10.0	18.0			
			As-cast	1500	28,000	In test					
			As-cast	1500	26,000	In test					
R402	18% Ni, 18% Cr, 0.2% C, 1.25% B, 5% Mo, 4% W	.2 C, 4 W	As-cast	1500	52,600	STTT	4.0	5.5	62,600	~ 45,000	30,200
			As-cast	1500	30,000	108	8.0	13.0			
			As-cast	1500	37,000	40.7	12.0	12.0			
			As-cast	1500	28,000	In test					
R403	18% Ni, 18% Cr, 0.3% C, 1.25% B, 5% Mo, 2% W	.3 C, 2 W	As-cast	1500	63,400	STTT	3.0	4.0	63,400		~ 30,000
			As-cast	1500	30,000	84.8	10.0	12.0			
			As-cast	1500	30,000	In test	120 hr				
			As-cast	1500	35,000	In test					
R405	18% Ni, 18% Cr, 0.3% C, 1.25% B, 5% Mo	Standard UM 233	2150-1-AC	1500	30,000	115	9.0	18.0			~ 30,300
			2150-5-AC	1500	30,000	46.2	10.0	12.0			~ 27,500
			2150-24-AC	1500	30,000	23.2	10.0	17.0			~ 25,500
			2150-1-AC	RT	76,700	STTT	3.0	4.5			
			2150-5-AC	RT	66,800	STTT	0.8	1.5			
R406	18% Ni, 18% Cr, 0.5% C, 1.25% B, 5% Mo, 1% W	.5 C UM 233 plus 1% W	2150-1-AC	1500	33,000	63.7	6.0	7.0			~ 31,500
			2150-5-AC	1500	33,000	33.7	4.0	7.5			~ 28,500
			2150-1-AC	RT	76,800	STTT	1.0	3.0			
			2150-5-AC	RT	68,000	STTT	0.7	(3)			
R407	18% Ni, 18% Cr, 0.2% C, 1.25% B, 3% Mo, 2% W	.2 C, 3 Mo, 2 W	As-cast	1500	30,000	35.7	12.0	20.5		33,500	27,300
			As-cast	1500	27,500	In test					
			As-cast	1500	25,000	282	14.0	28.0			
R408	18% Ni, 18% Cr, 0.2% C, 1.25% B, 3% Mo, 4% W	.2 C, 3 Mo, 4 W	As-cast	1500	61,500	STTT	7.0	8.5			~ 31,000
			As-cast	1500	33,000	37.3	7.0	11.0			
			As-cast	1500	30,000	In test					
			As-cast	RT	81,400	STTT	1.2	4.5			
R409	18% Ni, 18% Cr, 0.3% C, 1.25% B, 3% Mo, 2% W	.3 C, 3 Mo, 2 W	As-cast	1500	64,300	STTT	6.0	5.5			~ 30,00
			As-cast	1500	31,000	38.9	10.0	16.0			
			As-cast	1500	30,000	In test					
R410	14% Ni, 18% Cr, 0.3% C, 1.25% B, 5% Mo	14 Ni with .3 C, 1.25 B, 5 Mo UM 233	As-cast	1500	28,000	In test					
			As-cast	RT	75,500	STTT	1.0	1.5			
R411	14% Ni, 18% Cr, 0.5 C, 2.0% B, 5% Mo	14 Ni to complete series	As-cast	1500	In test	STTT					
			As-cast	RT	69,500	STTT	1.0	1.5			

(1) STTT - Short-time tensile test.  
 (2) Broke at gage mark or fillet.  
 (3) Reported values changed slightly from previous reports due to acquisition of additional data.  
 (4) Heat treatment specified by temperature, time at temperature, and cooling rate, as, for example:  
 2150-5-AC means 2150°F, 5 hr, air-cooled.

CODE:

- Points Based on Analyzed C and W
  - ⊙ Points Based on Analyzed C, Aim W
  - Points Based on Aim C, Aim W
- RXXX ← HEAT No.  
 30,000 ← 1500°F, 100hr. R.S.  
 5.0 ← 1500°F TENSILE ELONG.  
 1.0 ← R.-T. TENSILE ELONG.

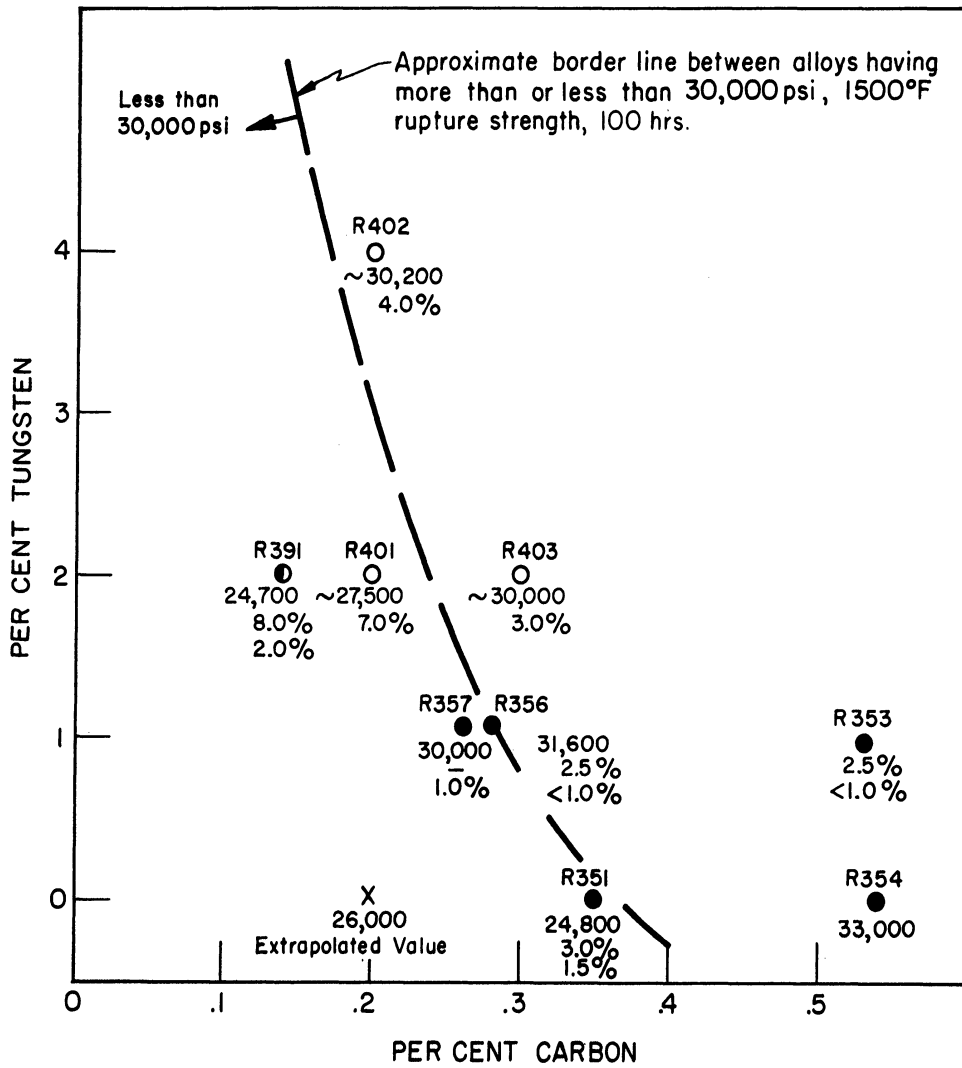


Fig. 1. Tungsten-carbon series.

• All heats contain 18% Ni, 18% Cr, 1.25% B, 5.0% Mo

TABLE I

## THERMAL-SHOCK TEST RESULTS

University of Michigan as-cast samples No. 372, 373, and 374.  
Standardized with HS 31 as-cast with couple attached.  
Temperature cycle - 800-1800°F/30 sec.

Specimen	Cycles Without Cracking	Cycles to Cracking	Cycles Without Failure	Cycles to Failure
<u>UM 233 Alloys</u>				
<u>Sample No. 372</u> - 18% Ni, 18% Cr, 0.3% C, 1.25% B, 5% Mo				
3	?	640	?	640
4	?	640	825	884
5	300	370	456	499
6	300	370	499	564
7	?	413	413	483
8	?	413	538	593
Avg	300	370	546	604
<u>Sample No. 373</u> - 18% Ni, 18% Cr, 0.2% C, 1.25% B, 5% Mo, 1% W				
3	?	305	370	444
5	305	370	571	634
6	142	340	340	403
7	198	273	427	507
8	222	294	355	426
Avg	217	321	414	544
<u>Sample No. 374</u> - 18% Ni, 18% Cr, 0.3% C, 2.0% B, 5% Mo				
3	?	221	365	439
4	221	293	354	438
5	260	328	501	673
6	200	260	516	688
7	325	393	542	625
8	325	393	464	542
Avg	266	334	457	567
<u>Haynes Stellite 31</u>				
Representative values	900	1300	1350	
<u>Ceramic Materials</u>				
Representative values	Very few	Very few	Very few	

Dimensional Changes

By placing a straight edge on back of the specimen a slight to no curvature was detected on all samples.

The bulging of the test edge is caused mainly by an adherent oxide scale. This scale could be noticed after approximately 200 cycles; after approximately 300 cycles the observation of cracking was difficult.



TABLE II

CARBON—MOLYBDENUM—TUNGSTEN SERIES  
 All heats contain 18% Ni, 18% Cr, 1.25% B

	0.2% C				0.3% C			
	3% Mo		5% Mo		3% Mo		5% Mo	
	Heat	100-hr R.S.	Heat	100-hr R.S.	Heat	100-hr R.S.	Heat	100-hr R.S.
0% W			26,000 Extrapolated value		R388	27,100	R351 R364 R369 R383	30,000 Avg
2% W	R407	27,300	R401	~27,500	R409	~30,000	R403	~30,000
4% W	R408	~31,000	R402	~30,200				

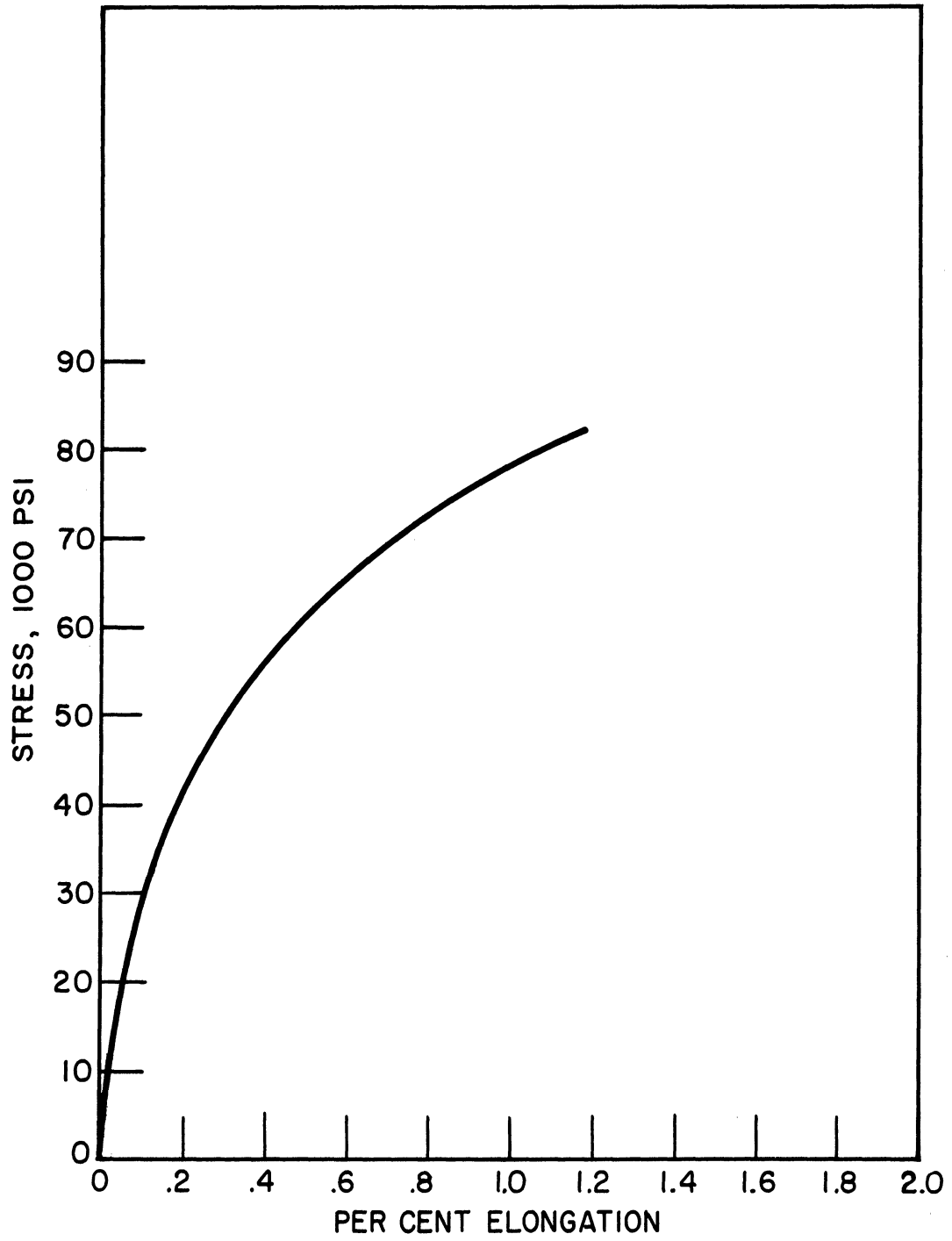


Fig. 2. Room-temperature stress-strain curve of heat R408.

TABLE III

## SUMMARY OF TOTAL BORON ANALYSES

Aim Boron	Heat No.	Laboratory				Other Elements Present*
		Colburn	Pitkin	Chrysler		
				Methyl Alcohol Method	New Ion- Exchange Method	
<u>Vacuum-Melted Heats</u>						
0	R310	.012				None
.1	R311	.198	.11, .12		.07	None
.5	R312	.333	.33		.30	None
	R344				.44	None
	R396				.43	None
1.0	R308	1.37, 1.51	.72		.88	None
	R395				.92	None
2.0	R309	1.79, 2.01	1.45, 1.50		1.53	None
<u>Argon-Melted UM 233 Heats</u>						
1.25	R336			.92	1.06	.5 C, 5 Mo
	R351			1.19	1.22	.3 C, 5 Mo
	R353				1.18	.5 C, 5 Mo, 1 W
	R354				1.26	.5 C, 5 Mo
	R356				1.24	.3 C, 5 Mo, 1 W
	R357				1.19	.2 C, 5 Mo, 1 W
	R364			.99	.99	.3 C, 5 Mo
	R366			.96	.96	.5 C, 5 Mo
	R369				1.02	.3 C, 5 Mo
	R383				1.13	.3 C, 5 Mo
	R391				1.09	.1 C, 5 Mo, 2 W
2.0	R338			1.72	1.81	10 Ni, .5 C, 5 Mo
	R349			1.99	1.87	.5 C, 2.5 Mo

\*Other elements present in addition to 18% Ni and 18% Cr, which were present on all heats listed.

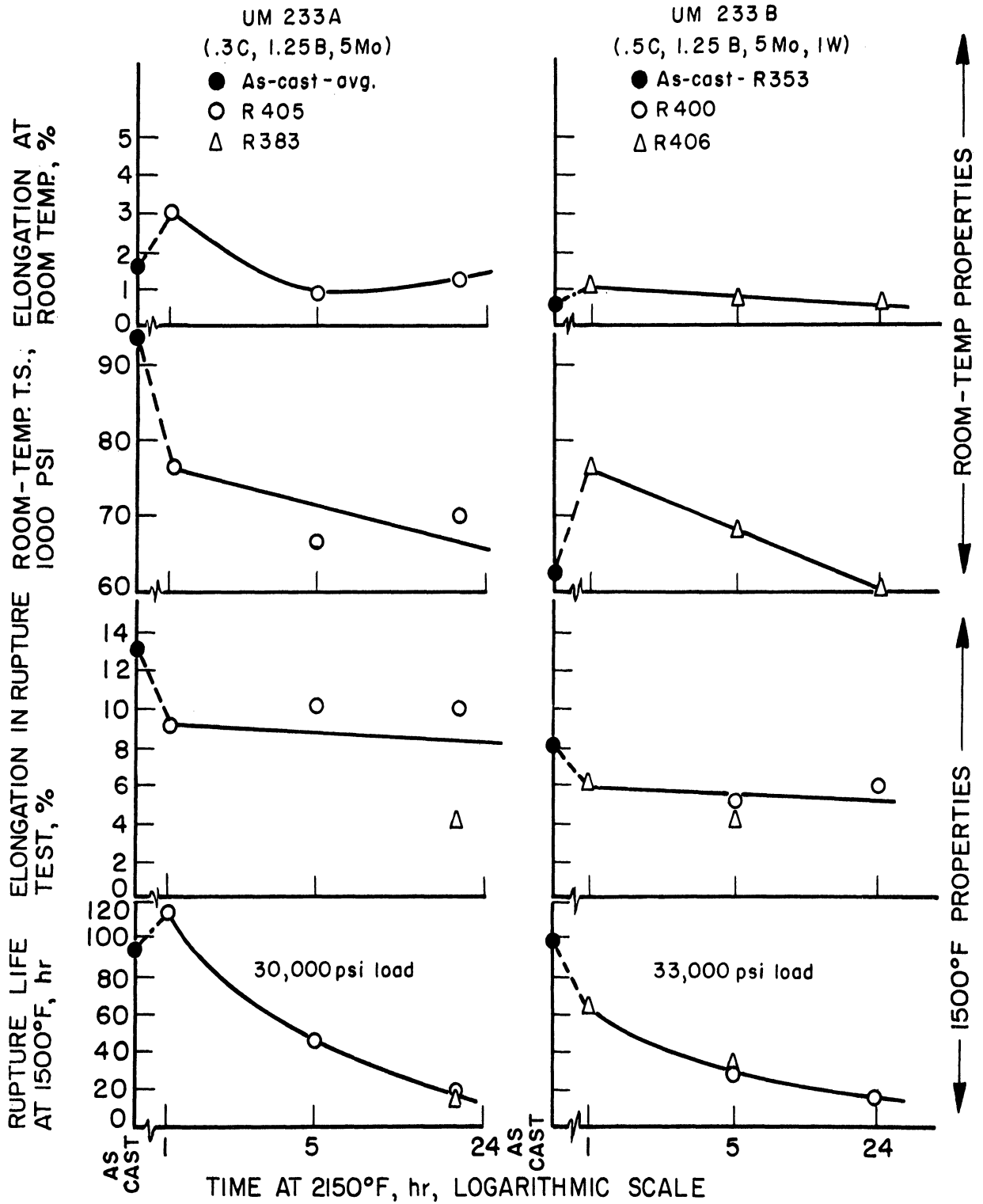


Fig. 3. Effect of 2150°F treatment on UM 233 alloys.

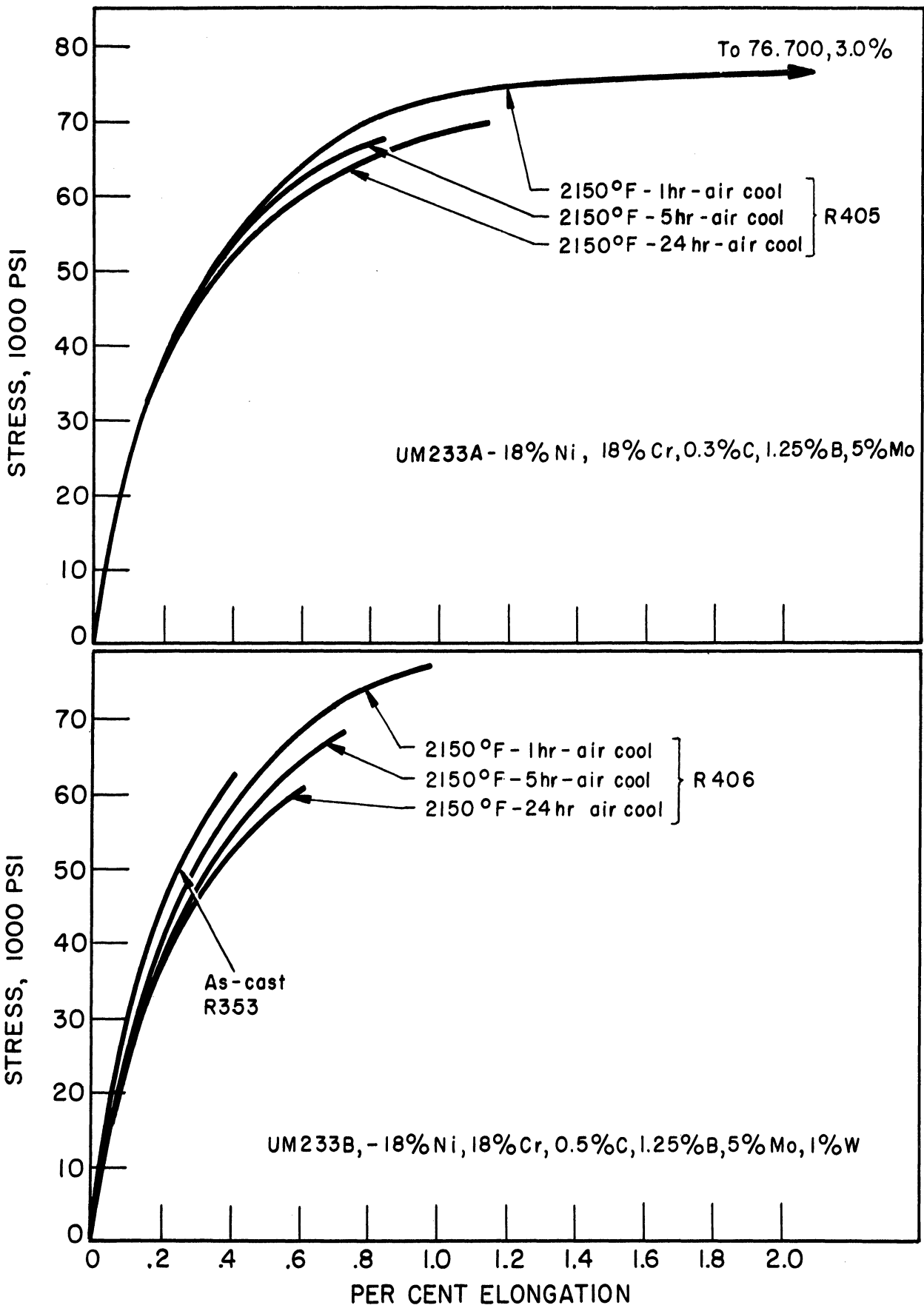


Fig. 4. Stress-strain curves of solution-treated alloys.