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COBALT-BASE ALLOY TESTING PROGRAM

R. B. Herchenroeder
Union Carbide Corporation

Prime Contract No. AT-30-2-GEN-16
Contract No. 103423
Manufacturing No. 99864

BROOKHAVEN NATIONAL LABORATORY
ASSOCIATED UNIVERSITY
UPTON, LONG ISLAND, NEW YORK 11973

April 21, 1967

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ABSTRACT

The purpose of the work conducted under the subject contract was to evaluate various cobalt-base alloy compositions as potential fast-reactor fuel cladding materials. Chromium containing and chromium free alloys were modified with varying amounts of molybdenum, columbium, vanadium, zirconium, titanium, and hafnium. These were evaluated with respect to hot-working characteristics, stress rupture properties at 1500°F, weldability, and ductility (via room temperature tensile tests). Results indicate that alloys containing titanium offer the best opportunity for achieving the strength goals of the program without detrimental side effects.

INTRODUCTION

The objective of the program was to test various alloy compositions as possible backup materials for those stainless steels which are currently being used as cladding materials in fast nuclear reactors. Major aspects of the program were to determine which elements added to a cobalt-base alloy would most likely cause the alloy to achieve a 100,000-hour rupture life at 8,000 psi and 1500°F in a molten sodium environment and yet remain ductile with a face centered cubic matrix. Further, it was desired to minimize the alloying content of elements which have high nuclear cross sections, elements which would tend to transmute to detrimental elements (such as boron transmuting to helium), and elements which might tend to dissolve in sodium.

Fifteen alloy compositions were designed for testing by the Stellite Division of Union Carbide Corporation in cooperation with personnel from Brookhaven National Laboratory.

An abbreviated statement of work which was to be performed under the contract follows: 40-pound heats of each of the compositions were to be melted in vacuum and cast into two ingots. The ingots were to be forged to slabs suitable for rolling to 5/8-inch thick plate, 3/8-inch thick plate, and to 0.062-inch thick sheet. Sheet from each of the heats was to be tested as follows:

- (a) Duplicate room temperature tensile tests.
- (b) Triplicate stress rupture tests of approximately 150 hours duration at 1500°F.
- (c) Duplicate bend tests of bead-on-plate welded material.
- (d) Metallographic examinations.

This work has now been completed and a description of it is contained herein.

EXPERIMENTATION AND DATA PRESENTATION

Fifteen alloys, as listed in Table I, were melted in a vacuum induction unit (see Figure 1) as per the melting sequence outlined in Table II. The chemical analysis points and analyses are listed in Table I. There were no difficulties encountered during melting of the alloys with the exception of a tendency of the charge material to bridge the crucible. When the bridge was broken loose and fell into the molten bath, violent boiling occurred which caused appreciable splatter of the molten metal and consequent loss of material. For this reason, it would be better to melt a slightly larger charge and use a larger crucible. This would allow a looser packing of the charge materials and would reduce the tendency of the charge materials to bridge the crucible during melting.

The ingots obtained from each heat (see Figure 2) were forged at 2150°F to nominally 3/4-inch thickness. Two replacement heats for Heats 66-587 and 66-591 were later melted and were forged at 2050°F with excellent results. No usable material could be obtained from the original heats because of poor hot-workability.

Results of the forging are illustrated in Figures 3 through 17. Comments on the forging of each heat are listed with the figures and are self-explanatory.

After grinding as many of the defects from the forged billets as possible, they were rolled at 2150°F to 5/8-inch thick plate. Figures 18 through 32 show the 5/8-inch thick slabs after rolling. Hardness measurements of the 5/8-inch thick plate were made and are listed in Table III.

Sheet material was also rolled at 2000°F to reduce the effect of oxidation; even so, the sheets had a rough surface. This was especially true of the material which did not contain chromium.* Originally, the sheet was annealed at 2000°F and water quenched, but because of short life in stress rupture tests, the sheet was reannealed for 15 minutes at 2150°F in an argon atmosphere and fan cooled in air.

The oxides accumulated during annealing at 2000°F were removed in a salt bath. The sheet was pickled to clean it. Those sheets which did not contain chromium were severely attacked and this resulted in an exceedingly rough surface. To rectify this, the material had to be cold-rolled to some thickness less than the desired 0.062-inch thickness. Some of the material had to be replaced by rolling a portion of the 3/8-inch thick plate to sheet. On this replacement material, the sheet was sandblasted to clean the surface. This sandblasted surface, prior to welding, was vapor blasted to further clean it.

* If future heats are produced, it would be advisable to finish the sheet by cold-rolling. This would provide a much smoother uniform surface.

Hardness measurements of 3/8-inch thick plate and the sheet which was annealed at 2150°F were made and the data are listed in Tables IV and V, respectively.

Tensile test specimens were machined from the sheet as per the drawing in Figure 33. These were tested at room temperature and the results are listed in Table VI and are shown graphically in Figure 34.

Stress rupture tests were performed at 1500°F at at least three stress levels using the tensile type test specimen. Results of the stress rupture tests are listed in Table VII.

The shoulder portions of each tested stress rupture specimen were bent about a radius equal to twice the thickness of the sheet to acquire an indication of the postaging ductility of the alloys. The rig used to bend the specimens is shown in Figure 54. The resulting bend angles are listed in Table VII along with the stress rupture data.

Hardness measurements of each tested rupture specimen were also made to determine if the exposure to temperature caused hardening to occur. These data are also listed in Table VII.

Table VIII lists the data obtained from the weld-bead-on-plate tests. Initially, all of the alloys were fusion welded using a Heliarc torch. All welding was done on a welding machine which controlled travel speed, amperage, voltage, and gas flow quite precisely, as noted.

Travel Speed -- 13.78 Inches Per Minute

Amperage -- 45

Volts -- 10.5

Gas Flow -- 30 Cubic Feet Per Hour on the Torch, 15 Cubic Feet Per Hour for Backup

With these conditions, centerline cracking developed in the welds of Heats 66-586, 66-588, and 66-589. This suggests that the travel speed and power input were too great. These alloys were subsequently rewelded using a hand torch, and no centerline cracking occurred. All samples were bent using the rig shown in Figure 54.

Metallographic examinations of as-cast ingots, plate, annealed sheet, and stress rupture tested sheet samples were made. Figure 39 through 53 are typical structures of these alloys.

DISCUSSION OF DATA

Examination of Table I, which is the chemical analyses of the heats, shows that the actual chemical analyses were very close to the aim points.

Tensile Data

Tensile tests of the material showed that Heat 66-593 had the highest strength, undoubtedly because of the five percent titanium included in the alloy.

Assuming, 40 percent tensile elongation is used as the minimum for good formability, many of the alloys fell below this mark. This could have been caused by the fact that many of the alloys seemed to be partially hexagonal close packed in structure.

In those alloys containing chromium, none of the alloying agents -- i.e., vanadium, molybdenum, columbium, zirconium, or titanium -- were particularly beneficial in causing increased yield strengths. All behaved essentially the same and did not improve the tensile strengths of the base composition represented by Heat 66-580. Perhaps, if the samples were pre-aged, some strengthening could be observed. This would particularly be expected for Heat 66-585, which contains titanium.

The tensile strengths of Heats 66-588, 66-589, and 66-590 did increase with increasing zirconium, but only Heat 66-590 was significantly superior to the base composition Heat 67-507.

Stress Rupture Data

In Table VII, stress rupture data are listed. Figures 35 through 38 are combined Larson-Miller and log-log plots of the stress rupture data. Figure 35 contains the stress rupture data of the better alloys for comparison purposes. Figures 36, 37, and 38 group the data from those alloys containing chromium, those without chromium and containing zirconium, and those containing other elements but also without chromium.

Data from sheet annealed at two conditions are presented. These conditions are: (a) ten minutes at a temperature of 2000°F, followed by a water quench; and (b) ten minutes at a temperature of 2000°F plus 15 minutes overall at 2150°F in an argon atmosphere, followed by a fan cool. As noted previously, the 2000°F temperature was used to reduce the oxidation which occurred during annealing.

The stress rupture life of material annealed at 2000°F was quite low. Based on experience with other cobalt-base alloys, viz., HAYNES alloy No. 25, HAYNES alloy No. 188, and HAYNES Developmental alloy No. 8187, which indicated that a higher annealing temperature would yield longer stress rup-

ture life. Samples of the sheet material were reannealed under the second annealing condition.

In general, the 150°F higher annealing temperature substantially increased the stress rupture life of the material, without decreasing elongation or postaging ductility. Exceptions to this generality were Heats 66-588 and 66-589, which contained zirconium. Whether or not the zirconium was influential in causing this difference could not be determined within the limited scope of the present program.

Inasmuch as the previously mentioned HAYNES alloys showed improvement by raising the annealing temperature 50°F from 2150°F to 2200°F, it is possible that further improvement in stress rupture life of these heats could be achieved by increasing the annealing temperature another 50°F. This would be a worthwhile experiment.

There are several possible explanations for the stress rupture life being improved by increasing the annealing temperature.

- (1) The higher temperature of 150°F increased the solid solubility of the matrix for solute atoms. Thus, during testing at 1500°F, the matrix was supersaturated to a greater degree and a difference in precipitation rate and resulting morphology of the precipitation would be expected.
- (2) Another possibility is that the solutioning of minor constituents altered the allotropic transformation of the base composition.
- (3) Still another possibility is that the higher annealing temperature promoted grain growth, which would reduce grain boundary sliding and would limit grain boundary diffusion.

The possible importance of the allotropic transformation with regard to the rupture properties of these alloys should not be overlooked. Many of the microstructures, Figures 39 through 53, have structures that are typical of that of hexagonal close packed cobalt. Since the transformation temperature from hexagonal close packed to face centered cubic of pure cobalt is approximately 783°F (417°C) and because the addition of refractory metals would tend to raise this transformation temperature, it is quite possible that some of the alloys were actually undergoing transformation during testing. Methods of determining whether or not allotropic transformations occurred in any of these alloys would be to perform either hot stage metallography or dilatometric studies.

Another factor which should be considered in evaluation of the stress rupture data is precipitation of secondary phases during testing. It is conceivable that the stress rupture life of the materials could be enhanced by preaging the sample prior to testing. But, determination of this was not within the realm of this program.

It is readily apparent (see Figure 35) that alloys containing titanium are superior in strength characteristics under the conditions of these tests. However, these alloys, containing titanium but no chromium, tend to have very low postaging ductility as exhibited by the bend test data in Table VII. These are Heats 66-592 and 66-593.

Based only on strength and postaging ductility, the best alloy appears to be HAYNES Experimental alloy No. EB 33166 as represented by Heat 66-586. This alloy has strength characteristics quite close to those of HAYNES alloy No. 25 from which the strength goals of this program were derived. Unlike HAYNES alloy No. 25, though, alloy No. EB 33166 did not seem to embrittle after aging, as can be seen by the bend test data in the table. Even after approximately 681 hours, the shoulders of the specimen bent 180 degrees about a radius which was twice the thickness of the sheet.

Possible detrimental features of the alloy are the relatively high tungsten content, 12 percent, and the relatively high nickel content, approximately 18 percent. The tungsten would be detrimental because of its high nuclear cross section, and it is possible that the nickel would be detrimental because of its solubility in the molten sodium environment for which these alloys are being surveyed. These aspects will be evaluated further at the Brookhaven National Laboratory.

Figure 36 shows that Heat 66-586 was the best composition of those containing chromium. This is true both of the material in the 2150°F annealed condition and in the 2000°F annealed condition. The second rated alloy of this group appears to be Heat 66-585 in the 2150°F annealed condition. The only alloying addition which did not seem to have a beneficial effect on the stress rupture properties was vanadium; nor were the strengthening effects of molybdenum, columbium, and three percent zirconium outstanding.

Figure 37 depicts the effects of nominally one-fourth, one, and three percent zirconium on the base composition of five percent tungsten, five percent nickel, three percent iron, balance cobalt. One percent zirconium appears to be most effective in strengthening the base.

Figure 38 illustrates the effects of molybdenum, titanium, and hafnium additions on the stress rupture properties of the base composition.

Titanium is the most effective strengthening agent in this base. However, all of the alloying additions tend to increase the strength of the base material under the conditions of the tests.

The data points presented in Figure 38 are open to conjecture. In Figure 38, the curve for Heat 66-593 is significantly steeper than that of Heat 66-594 which contains hafnium. If both were extrapolated to 100,000 hours, one of the goals of this program, Heat 66-594 would appear to be the superior alloy, considering only strength. Continuing this supposition, if one were to use the two test data points of Heats 67-506 at 15 and 13

ksi and then extrapolate to 100,000 hours, Heat 67-506 would closely approach the goal of 8 ksi at 1500°F for 100,000 hours. However, there is another data point for Heat 67-506 located at coordinate Larson-Miller parameter 43.1, 10 ksi which tends to discount this heat as a superior one.

The point of this conjecturing is that, considering the limited number of tests, their brevity, and in most cases the scatter of the data involved, it does not seem advisable to extrapolate these curves in order to determine the best alloy with regard to strength for 100,000 hours.

In summary of the data contained in Figures 35 through 38 and in Table VII, it appears that the base composition of Heat 66-586 is outstanding and that titanium additions to cobalt-base alloys significantly improve the stress rupture characteristics. However, titanium additions to cobalt-base alloys without chromium tend to cause embrittlement and loss of ductility after service at 1500°F.

Welding

Considering the many circumstances involved in these weld tests, namely, the very limited amount of welding performed on the alloys prior to welding for the tests, it is doubtful that any of the alloys could be discredited at this time on the basis of being unweldable. However, the low bend ductility (see Table VIII) of several of the samples -- for example, Heats 66-580, 66-582, 66-584, 66-588, and 66-591 -- indicates that the welding aspects should be more carefully examined before any additional work is done with these type compositions.

Metallographic Interpretations

Figures 39 through 53 show the microstructures of as-cast material, 5/8-inch thick plate, sheet annealed at 2150°F, and sheet after stress rupture testing. The specimen having the longest rupture life for each of the alloys was used in the last instance.

It is immediately evident that the desired face centered cubic matrix was not always achieved. Figure 39a appears to be a classical example of a hexagonal close packed cobalt structure. It is conceivable that allotropic transformations occurred in some of these alloys during testing operations, for example, in stress rupture testing or welding. Undoubtedly, if additional surveys are done on these types of alloys, additional effort will have to be expended to assure a face centered cubic matrix which is stable; the problems associated with dimensional change and metallurgical instability of a material undergoing an allotropic transformation would be too great to overcome.

A review of the microstructures in Figures 39 through 53 is in order.

Figure 39 shows in all instances (a, b, c, and d) what is apparently a solid solution hexagonal close packed structure. The small particles scattered throughout the matrix apparently are inclusions entrapped from the melt and in general, would not be expected to be detrimental. The dark areas along the grain boundaries in Figure 39d are areas of secondary fracture of the stress rupture specimen, indicating that the alloy fracture mode is intergranular rather than transgranular. There does not seem to be any precipitation in this base composition as a result of its exposure of 194 hours at 1500°F.

The microstructures in Figure 40 also indicate that this alloy, which has approximately five percent molybdenum, is also solid solution. Again, there are scattered particles throughout the matrix which are probably inclusions from the melt. Unlike the base composition, this alloy, after 302 hours at 1500°F, shows an intergranular precipitate of fine discrete particles. At the very low level of carbon, 0.02 weight percent, it is questionable but not completely impossible that these are a carbide of some type.

The structures of Heat 66-582, which contains nominally five percent columbium, are shown in Figure 41a through 41d. In the as-cast condition, there is an eutectic type secondary phase present. This phase is somewhat broken up and redistributed in the structure of the 5/8-inch thick plate in Figure 41b and apparently is completely dissolved in the sheet sample shown in Figure 41c. Figure 41c appears to be largely face centered cubic. This is somewhat surprising since columbium, being a body centered cubic material, should stabilize the hexagonal close packed allotrope of the base composition much the same as the molybdenum did. This is especially true since the base structure as exhibited in Figure 39 looks hexagonal. Unlike the two previous alloys discussed, alloy No. EB 32766, Heat 66-582, shows gross precipitation after 194 hours at 1500°F. The precipitation quite probably accounts for the samples low bend ductility bend angle of 40 degrees of the sample.

The microstructure of the vanadium bearing heat, shown in Figure 42, appears to be face centered cubic. The 5/8-inch thick plate has scattered particles of a second phase present, but these are largely dissolved in the sheet sample shown in Figure 42c. Little aging occurred during an exposure of 215 hours at 1500°F, as is reflected in the 180-degree bend angle of the material. Again, the fracture, as indicated by the secondary fractures along the grain boundaries, is intergranular rather than transgranular.

Heat 66-584, containing nominally three percent zirconium, has a gross amount of secondary phase in all conditions examined. The matrix appears to be a mixture of both face centered cubic and hexagonal, although it is difficult to prove by optical microscopy. The secondary phases were attacked readily by all of the etchants used, as can be seen by the rather dark appearance. In efforts to retain the secondary phase, the sample was not etched heavily enough to distinctly reveal the grain structure.

As will be seen in subsequent samples containing zirconium, this type of phase is typical of those samples which did contain zirconium. The amount of secondary phase correlates nicely with the level of zirconium in the sample. It is difficult to determine whether or not any fine precipitation occurred in the stress rupture sample. However, the bend angle of 80 to 90 degrees would indicate that some sort of precipitation did reduce the ductility of the material.

Apparently, a face centered cubic matrix was achieved in the sample containing titanium, Figure 44. In all cases - Figures 44a, 44b, 44c, and 44d -- the structures looked to be face centered cubic. There was some precipitation at the grain boundaries and slip planes of the stress rupture sample, Figure 44d. However, after 549 hours, the sample still maintained excellent ductility, having a bend angle of 180 degrees. This factor, along with the fact that titanium does significantly strengthen the material, further enhances the probability of meeting the long range goals of this program by the use of titanium additions, perhaps to the Heat 66-586 base.

This base, Heat 66-586, is shown in Figures 45a through 45d. The sample apparently is face centered cubic. By annealing at 2000°F as shown in Figure 44b, not all of the secondary phases are dissolved. The larger particles apparently are associated with the nominally one percent zirconium addition; however, the finer particles along the grain boundaries and slip planes could conceivably be carbides or some other type of inter-metallic. In the sheet sample annealed at 2150°F, all of these fine particles are dissolved. An interesting detail is that after 681 hours at 1500°F, the sample still had a 180-degree bend angle, even though there was gross precipitation throughout the matrix. The zirconium related phase coalesced from the elongated stringers shown in Figure 45c to the more or less globular particles shown in Figure 45d.

Again, the base composition represented by HAYNES Experimental alloy No. EB 33266, Heats 65-587 and 66-506, was intended to be face centered cubic in crystal structure and to serve as a basis for comparison of the other samples to which various amounts of other alloying agents were added. Unfortunately, Figure 46 indicates that alloy EB 33266 is largely hexagonal close packed in crystal structure and that, after aging for 112 hours at 1500°F (Figure 46d) there is little or no secondary precipitation.

Figures 47b and 47c also indicate largely hexagonal close packed crystal structures; however, Figure 47d may represent a complex hexagonal close packed-face centered cubic structure. A precipitate forms which forms uniformly throughout the structure during aging. Conceivably, a uniform precipitate like it could be used as a dispersion strengthening agent, if it were developed prior to subjecting samples to the test loads. Note, that in this structure (Figure 47), there are few secondary particles, in Figure 48, there is an increased amount of the larger elongated particles, and in Figure 49, there is a gross amount of elongated larger particles present. These are attributed to the zirconium additions. As mentioned

previously, this indicates that the selectively etched primary phase, shown in the cast structure, can be attributed to zirconium content also.

Figure 48 indicates that the structure should be largely hexagonal close packed and that a fairly heavy precipitation occurs during aging of the sample at 1500°F. Apparently, this aging does not detrimentally affect the bend ductility of the material.

The structure of alloy EB 33566 (Heat 66-590) shown in Figure 49 is somewhat similar to that of Figure 48 as mentioned previously except that the matrix appears to be a mixture of face centered cubic and hexagonal close packed and that the amount of the phase attributed to zirconium is much heavier. Further, the amount of precipitation in the stress rupture sample, Figure 49d, is much greater and the particles have an entirely different morphology; whereas, in Figure 48d, the particles arranged themselves in the more or less Widmanstätten structure, particles in Figure 49d did not.

Figures 50b and 50c substantiate the supposition that an allotropic transformation is occurring in at least some of these alloys. Figure 50c looks to be essentially face centered cubic in crystal structure, whereas Figure 50b represents a substantially hexagonal close packed one. This is something of an abnormality. One would expect that the 5/8-inch thick plate (Figure 50b) would cool substantially slower even with a water quench than the 1/16-inch thick sheet; therefore, the 5/8-inch thick plate should have a greater tendency to transform to hexagonal close pack than does the sheet. That is, the allotropic transformation would be more likely to be suppressed by the rapid cooling of the sheet than by the slower cooling of the 5/8-inch thick plate. Whether or not the stacking faults shown in Figure 50d are lined with a precipitate or not is difficult to discern at the magnifications used in this light microscopy study.

If additional heats of material containing molybdenum are to be evaluated, it would be beneficial to determine if there are precipitates present in this aged sample. A method of doing this is transmission electron microscopy, which would enable the phases present to be pinpointed and identified with regard to crystal structure.

The matrix of Heat 66-592, containing nominally two percent titanium, appears to be face centered cubic in the annealed condition as shown in Figures 51a, 51b, and 51c. Also, these structures indicate that all the secondary phases shown in Figure 51d can be dissolved. Actually, the microstructure of Figure 51d appears similar to a eutectoid steel. It would be interesting to determine whether or not the lamellar structure is of a eutectoid nature. However, this is beyond the scope of the investigation.

The series of micrographs, Figures 52a through 52d, are similar in nature to those of Figures 51a through 51d - with the exception that the cast structure shows a secondary phase and that very dark areas are shown

in Figure 52d. These are areas of quite dense lamellar phase which can be resolved by very high magnification. Both Heats 66-592 and 66-593 had low postaging ductility as exhibited by the bend angles of 30 and 25 degrees, respectively. Undoubtedly, this fact can be attributed to the eutectoid type structures shown.

Figure 53 represents the structure of the material containing one percent hafnium. There is a grain boundary phase present in the as-cast material, Figure 53a. This seems to dissolve during hot-working operations but reprecipitates as shown in the structure of the 5/8-inch plate, Figure 53b. The composition of the stringer like phase is unknown and its origin can only be of a speculative nature. Apparently, this matrix is face centered cubic in crystal structure, and appreciable portions of the secondary phase noted can be dissolved by annealing at a high enough temperature. This is shown by the structure (Figure 53c) of annealed sheet which indicates that the material is essentially solid solution with only a random distribution of secondary phase present. After exposure at 1500°F for 176 hours, there is a gross precipitation of some sort. It is possible that this precipitating phase could significantly strengthen the material providing it were caused to precipitate prior to the sample being subjected to a load. This could be beneficial because, even though the material did have large amounts of precipitation present, it did have quite adequate bend ductility -- 180 degrees after 176 hours at temperature.

CONCLUSIONS

- (1) None of the alloys tested has the desired strength level which would extrapolate to 8 ksi at 1500°F for 100,000 hours. Of those tested, the composition represented by Heat 66-586 has the best overall strength characteristics and postaging ductility. Questionable aspects of the alloy are its high tungsten and nickel contents because of considerations of nuclear cross section and corrosion by molten sodium, respectively.
- (2) The most influential strengthener was titanium. Titanium can significantly strengthen both chromium bearing and chromium free material but causes a gross lamellar precipitate in those samples without chromium. This results in exceptionally low postaging ductility. In the sample containing chromium, the alloy was strengthened, but little secondary precipitate was noted and the sample had excellent postaging ductility.
- (3) With the exception of vanadium, all of the elements added showed some indication of strengthening the base material.
- (4) Several of the alloys tested had either a hexagonal close packed matrix or a mixture of hexagonal close packed and face centered cubic crystal structure in the matrix. This undoubtedly affected the results of the testing program. If all of the samples had been face centered cubic in crystal structure, the data obtained probably would have been more uniform.
- (5) Increasing the annealing temperature of the samples from 2000°F to 2150°F significantly increased the stress rupture life of most of the samples with the two exceptions of the alloys which contained zirconium.
- (6) Several of the alloys incurred precipitation during service at 1500°F. It is possible that, had these alloys been preaged, the rupture life of these samples would have been materially increased.
- (7) It is not advisable to extrapolate the stress rupture curves shown to determine the best alloy with regard to strength for 100,000 hours because of the limited number of tests, their brevity, and in most instances, the scatter of data involved.
- (8) None of the alloys should be discredited at this time on the basis of unacceptable weldability.
- (9) No beneficial effects were determined in these alloys without chromium and they oxidized significantly.

RECOMMENDATIONS

- (1) A more thorough physical metallurgy study of the alloys of this program should be conducted to achieve better understanding of the metallurgical aspects. This would include the identification of the various precipitates which form and the determination of the probability of the probability of allotropic transformations in the alloys during testing.
- (2) In addition to conducting studies of corrosion in molten sodium and radiation effects, additional samples of the material should be preaged and tested to determine if precipitates observed in the microstructures would strengthen the material.
- (3) Selected samples should be annealed at a temperature of at least 2200°F and then tested to determine if this also would improve stress rupture life of the material.
- (4) If additional surveys of this program type are made, an additional effort should be made to obtain a face centered cubic matrix. If an alloy with desirable characteristics is found, compositional ranges which will assure a face centered cubic matrix should be determined.
- (5) At least two second generation alloys appear feasible and should be investigated. These are (a) the base composition of Heat 66-586 possibly without zirconium, but with titanium additions added to increase strength and (b) as an alternative in case the nickel and tungsten of the base composition of Heat 66-586 are too high, a sample containing titanium and molybdenum substituted in part for tungsten should be investigated.
- (6) Future heats should contain chromium.
- (7) The sheet from future heats should be finished by cold-rolling and then annealing to provide an improved surface finish on the material.

TABLE I

CHEMICAL ANALYSIS

HAYNES Experi- mental Alloy EB	Heat No.	Contract No.		Cr	W	Fe	C	Si	Ni	B	Al	Other
32566	66-580	1	● ●●	21.09 21.00	5.10 5.00	3.09 3.00	0.028 LAP	0.09 LAP	4.93 5.00	0.002 LAP	0.002 LAP	
32666	66-581	2	● ●●	20.99 21.00	5.08 5.00	3.11 3.00	0.022 LAP	0.08 LAP	5.07 5.00	0.002 LAP	0.02 LAP	5.10 Mo 5.00
32766	66-582	3	● ●●	20.98 21.00	5.02 5.00	3.03 3.00	0.012 LAP	0.03 LAP	5.09 5.00	0.002 LAP	0.08 LAP	4.96 Cb 5.00
32866	66-583	4	● ●●	21.18 21.00	4.97 5.00	2.96 3.00	0.012 LAP	0.04 LAP	5.02 5.00	0.002 LAP	0.14 LAP	4.94 V 5.00
32966	66-584	5	● ●●	21.03 21.03	4.94 5.00	2.99 3.00	0.022 LAP	0.02 LAP	5.11 5.00	0.002 LAP	0.13 LAP	3.04 Zr 3.00
33066	66-585	6	● ●●	21.01 21.00	4.99 5.00	3.00 3.00	0.012 LAP	0.04 LAP	5.10 5.00	0.005 LAP	0.10 LAP	1.79 Ti 2.00
33166	66-586	7	● ●●	19.11 19.00	11.93 12.00	2.06 2.00	0.017 LAP	0.04 LAP	17.93 18.00	0.005 LAP	0.11 LAP	0.96 Zr 1.00
33266	66-587	8*	● ●●		9.96 10.00	2.97 3.00	0.022 LAP	0.04 LAP	5.09 5.00	0.002 LAP	0.09 LAP	
33366	66-588	9	● ●●		10.07 10.00	3.06 3.00	0.016 LAP	0.03 LAP	5.05 5.00	0.002 LAP	0.11 LAP	0.29 Zr 0.25
33466	66-589	10	● ●●		10.12 10.00	3.11 3.00	0.021 LAP	0.04 LAP	5.01 5.00	0.002 LAP	0.12 LAP	1.06 Zr 1.00

TABLE I (Continued)

HAYNES Experi- mental Alloy EB	Heat No.	Contract No.		<u>Cr</u>	<u>W</u>	<u>Fe</u>	<u>C</u>	<u>Si</u>	<u>Ni</u>	<u>B</u>	<u>Al</u>	<u>Other</u>
33566	66-590	11	● ●●		10.15 10.00	3.10 3.00	0.016 LAP	0.05 LAP	4.99 5.00	<0.002 LAP	0.10 LAP	3.09 Zr 3.00
33666	66-591	12*	● ●●		4.97 5.00	3.12 3.00	0.016 LAP	0.04 LAP	5.10 5.00	<0.002 LAP	0.10 LAP	5.03 Mo 5.00
33766	66-592	13	● ●●		10.12 10.00	2.93 3.00	0.027 LAP	0.06 LAP	4.92 5.00	<0.002 LAP	0.08 LAP	2.06 Ti 2.00
33866	66-593	14	● ●●		10.03 10.00	3.04 3.00	0.022 LAP	0.06 LAP	5.06 5.00	<0.002 LAP	0.10 LAP	5.26 Ti 5.00
33966	66-594	15	● ●●		10.07 10.00	3.08 3.00	0.016 LAP	0.05 LAP	4.98 5.00	<0.002 LAP	0.09 LAP	0.92 Hf 1.00
33266	67-506	8	● ●●	0.26 LAP	9.37 10.00	2.48 3.00	0.003 LAP	0.03 LAP	5.70 5.00	<0.001 LAP	0.15 LAP	
33666	57-507	12	● ●●		4.52 5.00	3.08 3.00	0.001 LAP	0.02 LAP	4.90 5.00	<0.001 LAP	0.16 LAP	5.13 Mo 5.00

● Actual Analysis

●● Aim Points

* These heats have been remelted and forged.

TABLE II

MELTING SEQUENCE

Raw Material	Commercial materials were used: Electrolytic cobalt, nickel and iron, vacuum grade chromium, melting grade tungsten, sponge titanium, aluminum pellets, vanadium metal, hafnium metal and zirconium sponge.
Furnace	Fifty pound vacuum induction unit with Al_2O_3 crucible (See Figure 1).
Charging Sequence	All base elements were charged initially: Cobalt, chromium, nickel, tungsten, iron, vanadium, and molybdenum. Reactive elements were added late just prior to tapping; aluminum, titanium, zirconium, hafnium and columbium.
Melting Sequence	<ol style="list-style-type: none">1. Furnace charged.2. Shell evacuated.3. Power on.4. Temperature raised to 2600°F as measured with optical pyrometer*.5. Metal boil was controlled by alternately using vacuum pumps and blanking off.6. Metal held at 2600°F (optical measurement) until boil ceased.7. Aluminum was added to assure complete deoxidation.8. Vacuum shell was back filled with argon.9. Reactive elements were added and melted in.
Teeming	Heat was cast into: <ol style="list-style-type: none">1. One nominally 15-pound ingot (See Figure 2).2. One nominally 20-pound ingot.3. A mold for wet chemistry sample.4. A mold for X-ray and emission spectrographic samples.

* Actual temperature would be about 2750°F; optical pyrometer readings are about 150°F low because of sight glass in vacuum shell.

TABLE III
EXPERIMENTAL ALLOY HARDNESS
(5/8-Inch Plate)

<u>Heat No.</u>	<u>Rockwell A Hardness</u>				
66-580	58	60.5	61.5		
66-581	58.5	60	58		
66-582	69	70	71		
66-583	57	60	58.5		
66-584	61.5	62.5	63		
66-585	56	61.5	61	60	57
66-586	79	62	59	55	61
66-587	50.5	51.5	53		
66-588	78.5	55.5	57	55	53
66-589	57	55.5	58		
66-590	56	57	60		
66-591	56.5	55.5	55.5		
66-592	55	60	58	63.5	58
66-593	71	69.5	67.5		
66-594	54	53	56		
67-506	61	60.5	62		
67-507	65	64.5	62.5		

TABLE IV

EXPERIMENTAL ALLOY HARDNESS

(3/8-Inch Plate)

<u>Heat No.</u>	<u>Rockwell A Hardness</u>					
66-580	58	56	56			
66-581	59	61	61			
66-582	67	69	65			
66-583	57	60	62	58	59	61
66-584	62	65	62	63	63	
66-585	61	59	59	64	60	61
66-586	62	62	62	63	63	63
66-587				Scrap		
66-588	51	51	51			
66-589	58	60	58			
66-590	58	62	58			
66-591				Scrap		
66-592	58	61	61			
66-593	63	62	63			
66-594	58	56	59			
67-506	58	59	59			
67-507	65	67	67			

TABLE V

ANNEALED 16-MIL SHEET
ROCKWELL A HARDNESS

<u>Alloy No.</u>	<u>Heat No.</u>	<u>Hardness</u>		
EB-32566	66-580	61	61	60
EB-32666	66-581	64	63	64
EB-32766*	66-582	55	56	57
EB-32866	66-583	60	58	58
EB-32966	66-584	65	65	65
EB-33066	66-585	60	59	59
EB-33166	66-586	57	56	56
EB-33366	66-588	56	58	56
EB-33466	66-589	58	58	59
EB-33566	66-590	62	63	63
EB-33666	66-591	53	54	55
EB-33766	66-592	56	56	57
EB-33866**	66-593	63	62	62
EB-33966	66-594	57	55	56
EB-33266	67-506	57	58	58
EB-33666	67-507	59	57	58

* 0.038 sheet

** 0.097 sheet

TABLE VI
TENSILE DATA

<u>Alloy No.</u>	<u>Heat No.</u>	<u>Contract No.</u>	<u>Yield Strength, 0.2% Offset, KSI</u>	<u>Ultimate Strength, KSI</u>	<u>Elong., Percent</u>
EB-32566	66-580	1	60.8	132.5	23
			60.2	126.1	16
			• 60.5	129.3	20
EB-32666	66-581	2	61.0	127.4	35
			74.0	136.3	39
			• 67.5	131.5	37
EB-32766	66-582	3	72.2	132.2	64
			68.5	123.5	62
			• 70.0	127.5	63
EB-32866	66-583	4	55.8	120.2	46
			51.2	113.9	44
			• 53.5	117.0	45
EB-32966	66-584	5	67.5	131.7	19
			65.8	132.6	18
			• 66.0	132.5	18.5
EB-33066	66-585	6	54.4	117.9	38
			50.6	121.0	44
			• 52.5	119.4	41
EB-33166	66-586	7	51.9	125.6	61
			51.2	126.6	62
			• 51.4	126.0	61.5
EB-33366	66-588	9	39.0	100.4	39
			36.3	98.7	38
			• 37.5	99.0	38.5
EB-33466	66-589	10	41.4	108.3	51
			43.0	107.3	44
			• 42.2	107.5	47.5
EB-33566	66-590	11	73.3	128.1	26
			67.8	129.9	27
			• 70.5	129.0	26.5

TABLE VI (Continued)

<u>Alloy No.</u>	<u>Heat No.</u>	<u>Contract No.</u>	<u>Yield Strength, 0.2% Offset, KSI</u>	<u>Ultimate Strength, KSI</u>	<u>Elong., Percent</u>
EB-33666	66-591	12	40.5	86.2	13
			43.7	94.3	18
			● 42.1	90.5	15.5
EB-33766	66-592	13	56.1	114.0	77
			58.2	115.1	72
			● 57.5	114.5	74.5
EB-33866	66-593	14	87.2	159.7	41
EB-33966	66-594	15	41.7	108.5	59
			45.7	110.7	60
			● 43.7	109.6	59.5
EB-33266	67-506	8	36.8	102.0	48
			34.6	101.2	49
			● 35.7	101.5	48.5
EB-33666	67-507	12	47.7	111.3	38
			47.1	110.2	32
			● 47.4	110.7	35

All samples heat-treated ten minutes at 2000°F, water quenched, plus 15 minutes overall at 2150°F in argon atmosphere, fan cooled.

● Average value.

TABLE VII

STRESS RUPTURE DATA

<u>Alloy No.</u>	<u>Heat No.</u>	<u>Contract No.</u>	<u>Test Temp., °F</u>	<u>Stress, KSI</u>	<u>Life, Hours</u>	<u>Elong., Percent</u>	<u>Rockwell A Hardness</u>	<u>Bend Angle 2T Radius</u>	<u>Initial Condition of Sheet*</u>
EB-32566	66-580	1	1500	20	8.2	14	57, 58, 56	180, 180	A
			1500	17	24.5	23	57, 58, 56	180, 180	A
			1500	13	80.4	9	61, 58, 57	180, 180	A
			1500	13	130.4	13	50, 52, 41	180, 180	B
			1500	15	42.5	13	56, 55, 57	180, 180	B
			1500	10	194.5	2	57, 57, 56	180, 180	B
EB-32666	66-581	2	1500	20	20.2	30	64, 65, 65	180, 180	A
			1500	17	40.3	26	65, 66, 66	180, 155	A
			1500	13	136.3	10	66, 66, 65	180, 180	A
			1500	13	302	18	59, 58, 61	180, 180	B
			1500	17	27.9	14	62, 62, 58	180, 180	B
			1500	15	60.5	16	52, 56, 56	180, 180	B
EB-32766	66-582	3	1500	17	15.2	119	70, 71, 71	30, 35	A
			1500	17	46.0	64	66, 68, 68	40, 30	B
			1500	20	19.7	35	67, 66, 64	35, 40	B
			1500	13	194	69	68, 68, 69	40, 45	B
EB-32866	66-583	4	1500	15	31.1	15	54, 54, 56	180, 180	B
			1500	13	68.4	18	55, 55, 54	180, 180	B
			1500	10	215.8	18	55, 55, 56	180, 180	B
EB-32966	66-584	5	1500	20	7.7	75	64, 64, 64	80, 75	A
			1500	17	19.3	92	64, 66, 66	180, 165	A
			1500	13	81.9	56	65, 64, 65	60, 70	A
			1500	13	227.2	54	62, 62, 65	80, 90	B
			1500	17	30.9	64	62, 62, 61	80, 95	B
			1500	15	56	75	58, 59, 60	90, 100	B

TABLE VII (Continued)

<u>Alloy No.</u>	<u>Heat No.</u>	<u>Contract No.</u>	<u>Test Temp., °F</u>	<u>Stress, KSI</u>	<u>Life, Hours</u>	<u>Elong., Percent</u>	<u>Rockwell A Hardness</u>	<u>Bend Angle 2T Radius</u>	<u>Initial Condition of Sheet*</u>
EB-33066	66-585	6	1500	20	13.5	61	64, 65, 63	180, 180	A
			1500	17	28.0	75	64, 63, 62	180, 180	A
			1500	13	106.7	51	61, 62, 61	180, 180	A
			1500	20	19.5	42	56, 54, 55	180, 180	B
			1500	13	549.1	23	55, 55, 53	180, 180	B
			1500	15	134.1	41	58, 58, 55	180, 180	B
EB-33166	66-586	7	1500	20	27.5	61	63, 63, 62	180, 180	A
			1500	17	89.5	65	59, 59, 62	180, 180	A
			1500	15	132.1	60	62, 64, 62	180, 180	A
			1500	15	173.4	42	61, 62, 64	180, 180	A
			1500	20	72.9	19	52, 52, 51	180, 180	B
			1500	15	680.7	21	55, 54, 54	180, 180	B
EB-33266	66-587	8	S C R A P						
EB-33366	66-588	9	1500	17	14.6	12	55, 56, 53	180, 180	A
			1500	13	96.2**		49, 54, 55	180, 180	A
			1500	17	2.7	13	43, 43, 45	180, 180	B
			1500	17	1.3	6	38, 39, 42	180, 180	B
			1500	13	23	7	46, 48, 49	180, 180	B
EB-33466	66-589	10	1500	17	72.9	52	65, 56, 63	180, 180	A
			1500	13	96.0	31	50, 52, 52	180, 180	A
			1500	17	23.8	10	52, 53, 50	180, 180	B
			1500	20	6.0	10	48, 51, 52	180, 180	B
			1500	13	153.4	10	54, 55, 55	180, 180	B
EB-33566	66-590	11	1500	17	13.4	58	63, 65, 65	25, 40	A
			1500	13	39.2	68	60, 59, 59	30, 35	A
			1500	17	17.5	47	59, 60, 59	115, 80	B
			1500	13	85.0	55	59, 61, 61	80, 85	B
			1500	10	165.0	64	59, 59, 60	180, 180	B
			1500	10	165.7	66	59, 58, 58	180, 100	B

TABLE VII (Continued)

<u>Alloy No.</u>	<u>Heat No.</u>	<u>Contract No.</u>	<u>Test Temp., °F</u>	<u>Stress, KSI</u>	<u>Life, Hours</u>	<u>Elong., Percent</u>	<u>Rockwell A Hardness</u>	<u>Bend Angle 2T Radius</u>	<u>Initial Condition of Sheet*</u>
EB-33666	66-591	12	S C R A P						
EB-33766	66-592	13	1500	17	50.2	89	68, 67, 67	25, 30	A
			1500	13	84.4	55	68, 65, 68	30, 30	A
			1500	17	71.6	30	63, 63, 65	20, 20	B
			1500	15	116.4	24	53, 58, 58	30, 35	B
			1500	12	219.4	33	65, 65, 66		B
EB-33866	66-593	14	1500	17	159.7	23	64, 65, 68	30, 45	A
			1500	17	0.7	0.5	67, 67, 70	50, 30	B
			1500	13	257.2	37	57, 59, 60	40, 50	A
			1500	17	351.6	7	68, 67, 63	25, 35	B
			1500	20	0.3	2	70, 70, 71	50, 55	B
			1500	20	184.6	13	62, 63, 64	45, 45	B
EB-33966	66-594	15	1500	17	48.9	8	60, 59, 52	75, 90	A
			1500	13	116.9	20	52, 60, 56	90, 105	A
			1500	17	24.0	10	49, 51, 55	180, 180	B
			1500	13	176.3	10	52, 52, 53	180, 180	B
			1500	15	69.5	10	53, 55, 55	180, 180	B
EB-33266	67-506	8	1500	15	9.1	6	56, 58, 58	180, 180	B
			1500	10	112.3	8	54, 54, 55		B
			1500	13	45.5	7	48, 48, 47		
			1500	10	112.3	8	54, 54, 55	180, 180	B
EB-33666	67-507	12	1500	15	68.9	8	60, 58, 56	180, 180	B
			1500	13	97.7	10	56, 58, 58	180, 180	B
			1500	15	69.5	10	56, 57, 58		B
			1500	10	173.9	9	55, 55, 54	180, 180	B

* A = 10 minutes at temperature of 2000°F, water quench.

B = 10 minutes at temperature of 2000°F, water quench plus 15 minutes overall at 2150°F in argon atmosphere, fan cooled.

** Furnace burned out.

TABLE VIIIWELDED BEAD-ON-PLATE BEND TEST DATA

<u>Alloy No.</u>	<u>Heat No.</u>	<u>Contract No.</u>	<u>Bend Angle, Degrees</u>	<u>Sheet Thickness, Inch</u>
EB-32566	66-580	1	35, 40	0.060
EB-32666	66-581	2	180, 180	0.060
EB-32766	66-582	3	60, 40	0.060
EB-32866	66-583	4	180, 180	0.060
EB-32966	66-584	5	25, 25	0.070
EB-33066	66-585	6	180, 180	0.070
EB-33166	66-586	7	180, 95	0.056
EB-33266	66-587	8	Scrap	
EB-33366	66-588	9	10, 15	0.062
EB-33466	66-589	10	180, 170	0.070
EB-33566	66-590	11	180, 180	0.073
EB-33666	66-591	12	65, 50	0.055
EB-33766	66-592	13	180, 180	0.075
EB-33866	66-593	14	180, 180	0.100
EB-33966	66-594	15	180, 180	0.073
EB-33266	67-506	8	180, 110	0.070
EB-33666	67-507	12	115, 120	0.065

All material annealed ten minutes at 2000°F, water quenched, plus 15 minutes at 2150°F in argon atmosphere, fan cooled.

Figure 1: Vacuum, induction heated, melting furnace, 150 pounds capacity.

Figure 2: Typical ingots cast in program.

Figure 3
Neg. No. 21490

Forged billets HEA EB-32566
Ht. 66-580
21Cr, 5W, 3Fe, 66Co, 6Ni

Both ingots cracked slightly during first forging cycle; additional cracks appeared during subsequent cycles and original cracks propagated.

Lowest forge temp. 1850°F

Figure 4
Neg. No. 21491

Forged billets HEA EB-3266
Ht. 66-581
21Cr, 5W, 3Fe, 61Co, 5Ni

Small ingot forged fair; only small side cracks developed. Large ingot cracked during initial cycle with additional cracks developing during subsequent cycles. Ingots seemed hard.
Lowest forge temp. 1850°F

Figure 5
Neg. No. 21492

Forged billet HEA EB-32766
Ht. 66-582
21Cr, 5W, 3Fe, 61Co, 5Ni, 5Cb

Both ingots developed severe cracks.
Both ingots seemed exceptionally hard.

Lowest forge temp. 1830°F

Figure 6
Neg. No. 21493

Forged billets HEA EB-32866
Ht. 66-583
21Cr, 5W, 3Fe, 61Co, 5Ni, 5V

Ingots oxidized severely and scaled profusely. Cracks were initiated early, but were of a limited nature in the small ingot.

Lowest forge temp. 1820°F

Figure 7
Neg. No. 21494

Forged billets HEA EB-32966
Ht. 66-584
21Cr, 5W, 3Fe, 63Co, 5Ni, 3Zr

Ingots forged well.

Lowest forge temp. 1830°F

Figure 8
Neg. No. 21495

Forged billets HEA EB-33066
Ht. 66-585
21Cr, 5W, 3Fe, 64Co, 5Ni, 2Ti

Ingots were hard, but forged excellently.

Lowest forge temp. 1850°F

Figure 9
Neg. No. 21496

Forged billets HEA EB-33166
Ht. 66-586
19Cr, 12W, 2Fe, 48Co, 18Ni, 1Zr

Ingots forged well.

Lowest forge temp. 1850°F

Figure 10
Neg. No. 21497

Forged billets HEA EB-33266
Ht. 66-587
10W, 3Fe, 82Co, 5Ni

Oxides from ingots scaled fully during forging. Cracks developed early in sequence and extended well into ingot.

Lowest forge temp. 1840°F

Figure 11
Neg. No. 21498

Forged billets HEA EB-33366
Ht. 66-588
10W, 3Fe, 81.75Co, 5Ni, 0.25Zr

Ingots had shiny oxide. The small ingot developed a minor crack at the top. Forgeability was good in general.

Lowest forge temp. 1840°F

Figure 12
Neg. No. 21499

Forged billets HEA EB-33466
Ht. 66-589
10W, 3Fe, 81Co, 5Ni, 1Zr

Ingots forged well.

Lowest forge temp. 1830°F

Figure 13
Neg. No. 21500

Forged billets HEA EB-33566
Ht. 66-590
10W, 3Fe, 79Co, 5Ni, 3Zr

Ingots scaled profusely, but
forgeability was good.

Lowest forge temp. 1840°F

Figure 14
Neg. No. 21501

Forged billets HEA EB-33666
Ht. 66-591
5W, 3Fe, 82Co, 5Ni, 5Mo

Ingots seemed soft. Ingots
cracked severely; forgeability
was poor.

Figure 15
Neg. No. 21502

Forged billets HEA EB-33766
Ht. 66-592
10W, 3Fe, 80Co, 5Ni, 2Ti

Ingots forged well.

Lowest forge temp. 1800°F

Figure 16
Neg. No. 21503

Forged billets HEA EB-33866
Ht. 66-593
10W, 3Fe, 77Co, 5Ni, 5Ti

The small ingot developed a minor crack near its top. In general forgeability was good. Material seemed hard.

Lowest forge temp. 1820°F

Figure 17
Neg. No. 21504

Forged billets HEA EB-33966
Ht. 66-594
10W, 3Fe, 81Co, 5Ni, 1Hf

Ingots forged well but seemed hard.

Lowest forge temp. 1830°F

Figure 18
Neg. No. 21573

HEA EB-325
Ht. 66-580

5/8-inch thick hot rolled
plate

Figure 19
Neg. No. 21574

HEA EB-326
Ht. 66-581

5/8-inch thick hot rolled
plate

Figure 20
Neg. No. 21575

HEA EB-327
Ht. 66-582

5/8-inch thick hot rolled
plate

Figure 21
Neg. No. 21576

HEA EB-32866
Ht. 66-583

5/8-inch thick hot rolled
plate

Figure 22
Neg. No. 21577

HEA EB-32966
Ht. 66-584

5/8-inch thick hot rolled
plate

Figure 23
Neg. No. 21578

HEA EB-33066
Ht. 66-585

5/8-inch thick hot rolled
plate

Figure 24
Neg. No. 21579

HEA EB-33166
Ht. 66-586

5/8-inch thick hot rolled
plate

Figure 25
Neg. No. 21580

HEA EB-33266
Ht. 66-587

5/8-inch thick hot rolled
plate

Figure 26
Neg. No. 21581

HEA EB-33366
Ht. 66-588

5/8-inch thick hot rolled
plate

Figure 27
Neg. No. 21582

HEA EB-33466
Ht. 66-589

5/8-inch thick hot rolled
plate

Figure 28
Neg. No. 21583

HEA EB-33566
Ht. 66-590

5/8-inch thick hot rolled
plate

Figure 29
Neg. No. 21584

HEA EB-33666
Ht. 66-591

5/8-inch thick hot rolled
plate

Figure 30
Neg. No. 21585

HEA EB-33766
Ht. 66-592

5/8-inch thick hot rolled
plate

Figure 31
Neg. No. 21586

HEA EB-33866
Ht. 66-593

5/8-inch thick hot rolled
plate

Figure 32
Neg. No. 21587

HEA EB-33966
Ht. 66-594

5/8-inch thick hot rolled
plate

Figure 33

TENSILE & STRESS RUPTURE SPECIMEN

Figure 34

COMPARATIVE TENSILE DATA

Figure 35

STRESS-RUPTURE DATA

Figure 36

STRESS-RUPTURE DATA

Figure 37

STRESS-RUPTURE DATA

Figure 38

STRESS-RUPTURE DATA

METALLOGRAPHIC ETCHANTS

- (1) 80 milliliters of HCl plus 20 milliliters CH_3COOH .
- (2) 70 milliliters CH_3OH , 15 milliliters H_3PO_4 , 10 milliliters H_2SO_4 , 5 milliliters HCl, electrolytically.
- (3) 2 percent H_2CrO_4 .
- (4) 6 milliliters HCl plus 20 milliliters HNO_3 plus 50 milliliters saturated solution of CuCl_2 plus H_2O .

Metallographers: J. A. Martin and J. Simmons.

Neg. No. 22690 (a)

Neg. No. 22675 (b)

Neg. No. 22656 (c)

Neg. No. 22625 (d)

Figure 39
Magnification: 500X

HAYNES Experimental alloy No. EB-32566 - Heat No. 66-580 - 21Cr, 5W, 3Fe, 5Ni, Bal. Co.

- (a) As-cast. Etchant: 3
- (b) 5/8-inch thick plate; rolled at 2150°F; annealed 30 minutes at 2000°F and water quenched. Etchant: 1
- (c) 1/16-inch thick sheet; annealed 15 minutes at 2150°F; fan cooled. Etchant: 1
- (d) Annealed per (c) plus 194 hours at 1500°F; 180 degree bend angle. Etchant: 3

Neg. No. 22691 (a)

Neg. No. 22676 (b)

Neg. No. 22657 (c)

Neg. No. 22626 (d)

Figure 40
Magnification: 500X

HAYNES Experimental alloy No. EB 32666 - Heat No. 66-581 - 21Cr, 5W, 3Fe, 5Ni, 5Mo, Balance Co

- (a) As-cast. Etchant: 1
- (b) 5/8-inch thick plate; rolled at 2150°F; annealed 30 minutes at 2000°F and water quenched. Etchant: 1
- (c) 1/16-inch thick sheet; annealed 15 minutes at 2150°F; fan cooled; Etchant: 3
- (d) Annealed per (c) plus 302 hours at 1500°F; 180 degree bend angle; Etchant: 2

Neg. No. 22692 (a)

Neg. No. 22677 (b)

Neg. No. 22658 (c)

Neg. No. 22627 (d)

Figure 41
Magnification: 500X

HAYNES Experimental alloy No. EB-32766 - Heat No. 66-582 - 21Cr, 5W, 3Fe, 5Ni, 5Cb,
Balance Co.

- (a) As-cast. Etchant 2
- (b) 5/8-inch thick plate; rolled at 2150°F and annealed 30 minutes at 2000°F;
water quenched. Etchant 2
- (c) 1/32-inch thick sheet; cold rolled and annealed 15 minutes at 2100°F in
argon atmosphere; fan cooled; Etchant 3
- (d) 0.038-inch thick sheet; annealed per (c) plus 194 hours at 1500°F; 40 degree
bend angle; Etchant 2

Neg. No. 22693 (a)

Neg. No. 22678 (b)

Neg. No. 22659 (c)

Neg. No. 22628 (d)

Figure 42
Magnification: 500X

HAYNES Experimental alloy No. EB-32866 - Heat No. 66-583 - 21Cr, 5W, 3Fe, 5Ni, 5V, Balance Co

- (a) As cast. Etchant: 2
- (b) 5/8-inch thick plate; rolled at 2150°F and annealed 30 minutes at 2000°F and water quenched. Etchant: 2
- (c) 1/16-inch thick sheet; annealed 15 minutes at 2150°F; fan cooled; Etchant: 1
- (d) Annealed for (c) plus 215 hours at 1500°F; 180 degree bend angle. Etchant: 2

Neg. No. 22694 (a)

Neg. No. 22679 (b)

Neg. No. 22660 (c)

Neg. No. 22629 (d)

Figure 43
Magnification: 500X

HAYNES Experimental alloy No. EB-32966 - Heat 66-584 - 21Cr, 5W, 3Fe, 5Ni, 3Zr, Balance Co

- (a) As-cast. Etchant: 3
- (b) 5/8-inch thick plate; rolled at 2150°F and annealed 30 minutes at a temperature of 2000°F; water quenched. Etchant: 2
- (c) 1/16-inch thick sheet; annealed 15 minutes at 2150°F; fan cooled. Etchant: 2
- (d) Annealed per (c) plus 227 hours at 1500°F; 80-90 degree bend angle. Etchant: 2

Neg. No. 22695 (a)

Neg. No. 22680 (b)

Neg. No. 22661 (c)

Neg. No. 22030 (d)

Figure 44
Magnification: 500X

HAYNES Experimental alloy No. EB-33066 - Heat 66-585 - 21 Cr, 5W, 3Fe, 5Ni, 2Ti, Balance Co

- (a) As-cast. Etchant: 2
- (b) 5/8-inch thick plates; hot rolled at 2150°F and annealed 30 minutes at 2000°F. Etchant: 2
- (c) 1/16-inch thick sheet; annealed 15 minutes at 2150°F; fan cooled. Etchant: 1
- (d) Annealed per (c) plus 549 hours at 1500°F; 180 degree bend angle. Etchant: 2

Neg. No. 22696 (a)

Neg. No. 22681 (b)

Neg. No. 22662 (c)

Neg. No. 22631 (d)

Figure 45
Magnification: 500X

HAYNES Experimental alloy No. EB-33166 - Heat 66-586 - 19Cr, 12W, 2Fe, 18Ni, Balance Co, 1Zr

- (a) As-cast. Etchant: 3
- (b) 5/8-inch thick plate; rolled at 2150°F; annealed 30 minutes at 2000°F and water quenched. Etchant: 2
- (c) 1/16-inch thick sheet; annealed 15 minutes at 2150°F; fan cooled. Etchant: 1
- (d) Annealed per (c) plus 681 hours at 1500°F; 180 degree bend angle. Etchant 2

Neg. No. 22697 (a)

Neg. No. 22673 (b)

Neg. No. 22663 (c)

Neg. No. 22639 (d)

Figure 46
Magnification: 500X

HAYNES Experimental alloy No. EB-33266 - Heat 65-587 (a)- Heat 67-506 (b,c,d) - 10W, 3Fe, 5Ni, 82 Co.

- (a) As-cast. Etchant: 1
- (b) 5/8-inch thick plate; rolled at 2000°F; annealed 30 minutes at 2000°F and water quenched. Etchant: 4
- (c) 1/16-inch thick sheet; annealed 15 minutes at 2150°F; fan cooled. Etchant: 3
- (d) Annealed per (c) plus 112 hours at 1500°F; 180 degree bend angle. Etchant: 3

Neg. No. 22698 (a)

Neg. No. 22683 (b)

Neg. No. 22664 (c)

Neg. No. 22633 (d)

Figure 47
Magnification: 500X

HAYNES Experimental alloy No. EB-33366 - Heat 66-588 - 10W, 3Fe, 5Ni, 0.25Zr, Balance Co

- (a) As-cast. Etchant: 2
- (b) 5/8-inch thick plate; rolled at 2150°F; annealed 30 minutes at 2000°F and water quenched. Etchant: 2
- (c) 1/16-inch thick sheet; annealed 15 minutes at 2150°F; fan cooled. Etchant: 2
- (d) Annealed per (c) plus 23 hours at 1500°F; 180 degree bend angle. Etchant: 2

Neg. No. 22699 (a)

Neg. No. 22684 (b)

Neg. No. 22665 (c)

Neg. No. 22634 (d)

Figure 48
Magnification: 500X

HAYNES Experimental alloy No. EB-33466 - Heat No. 66-589 - 10W, 3Fe, 5Ni, 1Zr, Balance Co

- (a) As-cast. Etchant: 2
- (b) 5/8-inch thick plate; rolled at 2150°F; annealed 30 minutes at 2000°F and water quenched. Etchant: 2
- (c) 1/16-inch thick sheet; annealed 15 minutes at 2150°F; fan cooled. Etchant: 2
- (d) Annealed per (c) plus 153 hours at 1500°F; 180 degree bend angle. Etchant: 2

Neg. No. 22700 (a)

Neg. No. 22685 (b)

Neg. No. 22666 (c)

Neg. No. 22635 (d)

Figure 49
Magnification: 500X

HAYNES Experimental alloy No. EB-33566 - Heat 66-590 - 10W, 3Fe, 5Ni, 3Zr, Balance Co

- (a) As-cast. Etchant: 2
- (b) 5/8-inch thick plate; rolled at 2000°F; annealed 30 minutes at a temperature of 2000°F and water quenched. Etchant: 2
- (c) 1/16-inch thick sheet; annealed 15 minutes at 2150°F; fan cooled. Etchant: 2
- (d) Annealed per (c) plus 165 hours at 1500°F; 180 degree bend angle. Etchant: 2

Neg. No. 22701 (a)

Neg. No. 22674 (b)

Neg. No. 22667 (c)

Neg. No. 22640 (d)

Figure 50
Magnification: 500X

HAYNES Experimental alloy No. EB-33666 - Heat No. 66-591 (a) - Heat No. 67-507 (b,c,d) -
5W, 3Fe, 5Ni, 5Mo, Balance Co

- (a) As-Cast. Etchant: 2
- (b) 5/8-inch thick plate; rolled at 2000°F; annealed 30 minutes at a temperature of 2000°F; water quenched. Etchant: 4
- (c) 1/16-inch thick sheet; annealed 15 minutes 2150°F; fan cooled. Etchant: 1
- (d) Annealed per (c) plus 173 hours 1500°F; 180 degree bend. Etchant: 2

Neg. No. 22702 (a)

Neg. No. 22687 (b)

Neg. No. 22668 (c)

Neg. No. 22636 (d)

Figure 51
Magnification: 500X

HAYNES Experimental alloy No. EB-33766 - Heat 66-592 - 10W, 3Fe, 5Ni, 2Ti, Balance Co

- (a) As-Cast. Etchant: 2
- (b) 5/8-inch thick plate; rolled at 2150°F; annealed 30 minutes at a temperature of 2000°F; water quenched. Etchant: 2
- (c) 1/16-inch thick sheet; annealed 15 minutes at 2150°F; fan cooled. Etchant: 2
- (d) Annealed per (c) plus 116 hours 1500°F; 30 degree bend angle. Etchant: 2

Neg. No. 22703 (a)

Neg. No. 22688 (b)

Neg. No. 22669 (c)

Neg. No. 22637 (d)

Figure 52
Magnification: 500X

HAYNES Experimental alloy No. EB-33866 - Heat 66-593 - 10W, 3Fe, 5Ni, 5Ti, Balance Co

- (a) As-cast. Etchant: 2
- (b) 5/8-inch thick plate; rolled at 2150°F; annealed 30 minutes at a temperature of 2000°F; water quenched. Etchant: 2
- (c) 1/16-inch thick sheet; annealed 15 minutes at 2150°F; fan cooled. Etchant: 2
- (d) Annealed per (c) plus 352 hours at 1500°F; 25 degree bend angle. Etchant: 2

Neg. No. 22704 (a)

Neg. No. 22689 (b)

Neg. No. 22670 (c)

Neg. No. 22638 (d)

Figure 53
Magnification: 500X

HAYNES Experimental alloy No. EB-33966 - Heat 66-594 - 10W, 3Fe, 5Ni, 1Hf, Balance Co

- (a) As-cast. Etchant: 2
- (b) 5/8-inch thick plate; rolled at 2150°F; annealed 30 minutes at a temperature of 2000°F; water quenched. Etchant: 2
- (c) 1/16-inch thick sheet; annealed 15 minutes at 2150°F; fan cooled. Etchant: 2
- (d) Annealed per (c) plus 176 hours at 1500°F; 180 degree bend angle. Etchant: 2

Figure 54
Negative No. C-4979

Figure 54: Bend Test Rig.