# PHASE STABILITY OF TWO INVESTMENT CAST HIGH CHROMIUM NICKEL BASE ALLOYS

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## $A\,bstract$

Investigations were made of the changes occurring in the microstructural characteristics of Udimet 710 (18% Cr) and IN 738X (16% Cr) nickel-base alloys upon exposure to temperatures in the range of 1500 to 2200 F. These two alloys have been developed to provide an exceptional balance of high temperature strength and sulfidation resistance. Data from hot corrosion and stress rupture testing is reviewed. Both materials were initially investigated after being solution treated and aged, following a simulated aluminum diffusion coating cycle. Elevated temperature exposures were subsequently made of heat treated specimens for periods of time to approximately 2000 hours.

Optical and electron microscope techniques were used to observe the effects of the exposures on the solution treated and aged microstructure. Precipitated phases were extracted electrolytically and identified by X-ray diffraction after a variety of time-temperature exposures. Tensile and stress rupture tests were made on both alloys after the long time exposures at 1600 F and results compared with similar tests on the as-heat treated material. All experimental results are related to calculated electron vacancy numbers for the specific compositions evaluated and for the ranges of chemical composition proposed for the two alloys.

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### Introduction

The recent development of nickel-base alloys which have equivalent high temperature strength to alloys such as Inco 713C and Udimet 700 and, at the same time, have a high level of resistance to hot-corrosion represent a significant metallurgical contribution to gas turbine technology. These new alloys, which include Udimet 710 (1), IN-738X (2) and MAR-M 421 (3), promise improvement in both the reliability and economy of turbine engine operation. Development testing of certain of these new alloys has indicated that because of their improved resistance to hot-corrosion, their use should virtually end the premature removal of turbine blades from aircraft engine service for reasons of severe sulfidation-oxidation attack. Hot corrosion evaluations of these alloys have been made in the Allison Materials Laboratories and subsequently in T56 turbine engine tests. The improvement in corrosion-resistance of two of these alloys over Inco 713C is shown in Figure 1 by photographs of first stage turbine blades after engine testing. The engine tests were run at an average turbine inlet temperature of 1780 F with the continuous ingestion of synthetic sea water (ASTM Specification D1141-52) into the combustion section; ratio of synthetic sea salt to air was 0.75 ppm.

In addition to the desirable combination of high hot strength and resistance to hot-corrosion, there has been another development objective for these new nickel-base alloy compositions to meet; this is that they should have little or no tendency to form sigma or other objectionable intermetallic phases upon long time elevated temperature exposure. After the initial assessment of sulfidation resistance and mechanical properties of these alloys by the Allison Materials Laboratories, a study was made of the microstructural instabilities caused by elevated temperature exposures. This paper will report the results of these studies on the alloys Udimet 710 and IN-738X.

# Experimental Procedure

# Material and Processing Description

The chemical analysis of the Udimet 710 and the IN-738X materials used in this investigation is shown in Table 1. Both alloys were obtained in the form of 0.252 diameter cast to size stress rupture test bars; Udimet 710 from the Central Foundry Division of General Motors Corporation, Bedford, Indiana and the IN-738X alloy from Stellite Works, Material Systems Division of Union Carbide Corporation, Kokomo, Indiana.

Each alloy was individually processed through a thermal cycle simulating the first step of an aluminum diffusion coating process. This consisted of packing the specimens in an Al<sub>2</sub>O<sub>3</sub> filled retort and heating it to 1325 F under a flowing argon atmosphere. The retort was removed from the furnace when the temperature at the center of the pack reached 1325 F and was allowed to cool to room temperature before the specimens were removed. Each alloy was then further solution treated and aged as shown in Table 2. All of the specimens were then machined to remove the oxide layer and adjacent alloy depleted zone that was formed during the heat treatment. Fourteen bars were then cut transversely at the midpoint of the reduced section so that each bar provided two similar specimens for the thermal exposure part of the program. Ten full size specimens were preserved intact for mechanical testing after being exposed for 2000 hours at 1600 F.

Table 3 details the thermal exposure schedule which was followed in determining the phase stability of these two alloys. After the individual specimens were removed from the thermal environment a 1/4 inch long section was cut from the reduced section and mounted for metallographic analysis.

### Metallographic Specimen Preparation

These 1/4-in. long specimens were mounted in Bakelite and then surface ground to remove from 0.030 to 0.050 inch. They were then wet ground on 600 grit abrasive paper and polished on a one micron diamond lapping wheel. Final polishing was conducted electrolytically in a one part  $H_2SO_4$  to six parts methanol solution at 22 volts for approximately three seconds. Specimens were etched for both light and electron microscopic examination in a diluted (25%) solution of Marbles reagent (4 gm of  $CuSO_4$  and 20 ml of HCl diluted with 20 ml of  $H_2O$ ).

Examination of each material and condition on the electron microscope was conducted on single stage collodion (2%) replicas shadowed at approximately a 30° angle with chromium. Light photomicrographs were prepared at 500 times size and the electron micrographs exposed at 2500 times size and enlarged (2X) to 5000 times size for publication.

#### Phase Identification

The remainder of the specimen was then machined to remove any oxides and accompanying alloy depletion zones that had formed as a result of the high temperature exposures. This specimen was then used for the phase extraction and X-ray diffraction identification of phases in

the alloy; this work was done by Dr. John F. Radavich, Micro-Met Laboratories, Inc., West Lafayette, Indiana. Electrolytic extractions were carried out in a 10% HCl-Methanol solution. Operating parameters were 5 volts at 1/2 amps for 2-4 hours. Specimens were removed from the electrolyte at frequent intervals and scraped with a Teflon rod into fresh alcohol to remove the residue from the surface. A sufficient amount of residue was collected from each sample for the X-ray analysis. The X-ray analysis was conducted on a Phillips X-ray diffractometer equipped with a diffracted beam monochromater. CuK $_{\alpha}$  radiation was used and the diffractometer was run at 2°/min. The unit was scanned across the 2 $\Theta$  range where the most intense peak of the particular phase was located.

### Mechanical Testing

Both room temperature and elevated temperature tensile tests were conducted on a Riehle 60,000 pound testing machine at a rate of 0.035 in. per minute. Elevated temperature tensile specimens were allowed to soak at the test temperature 15-30 minutes before any load was applied. The stress rupture tests were conducted on Arcweld creep rupture testing machines having a lever arm ratio of 20:1. Specimens were instrumented with 2 Chromel-Alumel thermocouples attached near the center of the reduced section. Data on mechanical tests conducted after 2000 hour exposure at 1600 F was compared to data obtained on similar material which had not received the long time 1600 F exposure.

### Results of Structural and Property Studies

### Metallographic Analysis

Udimet 710 Alloy. The microstructure of Udimet 710 alloy in the as heat treated condition is shown in Figure 2. In this condition the structure consists of three sizes of gamma prime in a gamma matrix with the MC type carbide occurring in both intragranular as well as intergranular sites while a smaller  $M_{23}C_6$  is found in the grain boundaries.

1500 F

Figures 3 and 4 show the microstructure after 1000 and 2000 hours exposures. The primary effects of exposure at this temperature are (1) the elimination of the fine gamma prime phase from the microstructure, (2) the formation of more  $M_{23}C_6$  phase in the grain boundaries, and (3) the rounding of gamma prime phase. There is very little

growth of the intermediate gamma prime phase from the heat treated condition and even an exposure of 2000 hours at this temperature does not cause any change in size from the 500 hour exposure.

1600 F

Microstructures of the 1000 and 2000 hour exposures at this temperature are shown in Figures 5 and 6. The fine gamma prime phase is also eliminated during these exposures but the main effect of time at this temperature is shown by the rounding and growth of the intermediate gamma prime phase and the obvious increase in both the size and amount of the grain boundary carbide phase. There is some evidence of the formation of a rod like phase usually associated with the carbides and the beginning of grain boundary encapsulation with gamma prime can be seen after the 2000 hour exposure.

1700 F

The chain like carbide phase and the gamma prime encapsulation shown in Figures 7 and 8 are the most pronounced features of exposure at this temperature. The gamma prime phase can also be seen to increase in size as a function of time and the rounding of this phase is evident after 500 hours. Some isolated particles of the rod like phase were found after the 2000 hour exposure. The in-grain carbide phase appears to be smaller and more rectangular in shape than was found in the heat treated microstructure.

1800 F

Exposures at this temperature cause a more rapid increase in the size of the gamma prime, as would be expected, and a larger and more continuous capsule of gamma prime around both the grain boundaries and the in-grain carbide phase. The in-grain carbides are relatively small and rectangular in shape as was found in the 1700 F specimens. These features are seen in Figures 9 and 10.

1900 F

Exposures were made for 10, 100, and 500 hours at this temperature. 500 hour exposure microstructure is shown in Figure 11. The primary effect of time at this temperature appears to be the partial solutioning of some of the gamma prime, which is evidenced by the presence of the fine (less than 0.05 micron) reprecipitated gamma prime in the

matrix. Additional changes are growth of the balance of the gamma prime, continued formation of the M<sub>23</sub>C<sub>6</sub> phase and encapsulation of the grain boundaries with gamma prime.

2000 F

The microstructure after 500 hours at this temperature is shown in Figure 12. The increase in the amount of fine reprecipitated gamma prime in the matrix is evidence of the increased solutioning of gamma prime phase. The massive MC carbide phase is found at both intergranular and intragranular sites. There was no evidence of any  $M_{23}C_6$  phase in the microstructure after 100 hours but there was a small amount left after the 10 hour exposure.

2100 F

After 10 hours at this temperature the gamma prime is about 95% solutioned as shown in Figure 13. There was still some M<sub>23</sub>C<sub>6</sub> grain boundary carbide phase present after 1 hour but none could be found after the 10 hour exposure. The solutioning of the gamma prime was only about 70% complete after the 1 hour exposure.

2200 F

Total solutioning of the gamma prime phase occurred after I hour at this temperature. The microstructure after 10 hours exposure is shown in Figure 14. There is no visible effect of holding the sample for 10 hours when comparing the microstructure to the 1 hour exposure. The carbide phase in both cases is the massive MC type which appears to have, in many cases, a mottled surface.

IN-738X Alloy. The response of the microstructure of IN-738X alloy to the times and temperatures shown in Table 3 is depicted in Figures 15 through 27. The microstructure of the alloy in the as heat treated condition is shown in Figure 15. There are two sizes of gamma prime in this material as heat treated. The finer of the two is about 0.1 microns while the larger varies from 1.5 to 2.5 microns. The finer has a more rounded shape while the larger is more angular. Both the MC and the M<sub>23</sub>C<sub>6</sub> carbides are found in the alloy, the latter being found primarily at the grain boundaries. A variation in both the size and the shape of the gamma prime phase has been noted in a limited number of bars examined in the as heat treated condition.

### 1500 F

Exposure at this temperature for 1000 and 2000 hours is shown in Figures 16 and 17. Most of the finer gamma prime phase is gone even after 500 hours and the phase is found to become more rounded in appearance as the time is increased. The amount of M<sub>23</sub>C<sub>6</sub> is also increasing at the grain boundaries. There is still much of the MC carbide present in both intergranular and intragranular locations. A rod like phase is found in small amounts near carbides in this 2000 hour sample.

#### 1600 F

The effect of exposure at this temperature is very much like the effect at 1500 F. This is shown in Figures 18 and 19. The finer gamma prime phase is disappearing and the balance of the gamma prime is increasing in size and becoming more rounded. The amount and size of the M<sub>236</sub> phase found at the grain boundaries is also increasing. The start of a grain boundary gamma prime encapsulation is also evident. The large MC phase is still found at grain boundaries and in-grain sites.

#### 1700 F

The continued growth of the gamma prime phase with time at this temperature is shown in Figures 30 and 21. The grain boundaries are becoming more completely encapsulated with gamma prime and the intergranular carbide is becoming more continuous with the M<sub>23</sub>C<sub>6</sub> phase increasing in amount. The MC carbide phase now appears to be primarily located at in-grain sites. A small rectangular phase can be found in many locations but most frequently associated with the MC type carbide.

#### 1800 F

Figures 22 and 23 show the microstructure of this alloy after 1000 and 2000 hours, respectively. The most obvious features are the increased size of the gamma prime and the wide capsule of gamma prime surrounding the grain boundaries. Close examination also reveals a very fine reprecipitated gamma prime appearing in the gamma matrix after the 2000 hour exposure. The grain boundaries still contain both types of carbides while the in-grain carbides appear to be only the MC type.

1900 F

Even after ten hours exposure at this temperature the fine gamma prime phase can be found in the background. The appearance of the microstructure after 500 hours is shown in Figure 24. The size of the rest of the gamma prime is larger than as heat treated but the amount is much less. There appears to be only the MC phase present after the 500 hour exposure.

2000 F

Increased solutioning of the gamma prime phase and precipitation of the fine gamma prime\*can be seen to occur with time at this temperature. Figure 25 shows the microstructure of the alloy after 500 hours exposure. The MC type carbide is the only one found in this and other shorter time specimens exposed at this temperature.

2100 F

Complete solutioning of the gamma prime has not occurred even after 10 hours at this temperature. This is shown in Figure 26. As would be expected, only the MC type carbide phase is found in these samples.

2200 F

Figure 27 shows the microstructure after 10 hours at this temperature. Complete solutioning of the gamma prime phase accompanied by reprecipitation of a fine, evenly dispersed gamma prime phase was found after a 1 hour exposure. The massive MC phase is present at both intergranular and intragranular sites.

Phase Extraction and Identification

The results of the X-ray diffraction analysis of the electrolytically extracted phases from the two alloys are shown in Tables 4 and 5. The values shown in the tables are the measured peak heights of the (333) peak for the  $M_{23}C_6$  phase and the (111) peak for the MC phase. The ratio of these values, shown in the table, is also a ratio of the relative amounts of the two phases in the particular specimen.

\*upon cooling from 2000°F

Mys.

Udimet 710 Alloy. Examination of this ratio for the samples analyzed in the Udimet 710 alloy show that this alloy forms M23C6 at the expense of MC in the temperature range 1500-1900 F. At any specific temperature in the range of 1500-1800 F the amount of  $M_{23}^{C}$ is also seen to increase with the amount of time at that temperature, at least up to the maximum time of 2000 hours used in this study. At temperatures above 1900 F there is no  $M_{23}C_6$  phase found indicating that even that which was present in the heat treated alloy has now gone completely into solution and the only carbide left is the MC type which was found to have a lattice parameter of 4.31 A°. This phase has also been reported to have the composition  $Ti_{0.86}Mo_{0.10}W_{0.04}C$  and a lattice parameter of 4.322 A°.(4) Although the exact solution temperature of the M23C6 phase has not been established in the laboratory, the extraction data indicate that it is between 1900 and 2000 F. The possible presence of other phases were indicated in this alloy in the heat treated condition as well as after exposures at 1500, 1600, 1700, and 1800 F. In some cases the phase could be identified as borides and/or sigma phase but in the case of the 1700 and 1800 F exposures for 1000 hours the exact identification of the phase was impossible.

IN-738X Alloy. The data for IN-738X alloy also shows the tendency for the formation of the M23C6 phase as a function of exposure temperature but to a much lesser degree than was seen in Udimet 710 material. The maximum amount of  $M_{23}^{C}_{6}$  came after an exposure of 2000 hours at 1700 F. The maximum temperature at which the  $M_{23}^{C}_{6}$  was found is 1800 F. The tendency for the formation of M<sub>2.3</sub>C<sub>6</sub> to occur with time at any temperature is also noted in this alloy. However, the most significant change is seen to occur during the first 1000 hours of exposure at any temperature. The data indicates that the solution temperature of the M<sub>22</sub>C<sub>4</sub> phase occurs between 1800 and 1900 F. The MC carbide, extracted after 10 hour exposure at 2200 F, was found to have a lattice parameter of 4.36 A°, slightly larger than that found for the MC phase in the Udimet 710 alloy. The only specimen in this alloy in which there was even an indication of the presence of a secondary phase was the 1500 F/1000 hour exposure where there were some X-ray diffraction lines present which could have been from a sigma phase. However, neither the intensity nor the number of diffraction peaks was great enough to make a positive identification of the phase.

In no case for either alloy was the presence of significant amounts of sigma phase or other secondary phases noted, indicating the high degree of microstructural stability of these alloys when exposed to temperatures and times where the formation of these phases have been noted in other nickel base alloy compositions.

### Mechanical Testing

Uniaxial tensile tests and elevated temperature stress rupture tests were conducted on both alloys in the heat treated condition and after 2000 hours exposure at 1600 F. The results of these tests are shown in Tables 6 and 7.

Udimet 710 Alloy. Examination of the mechanical test data for Udimet 710 alloy clearly shows the effect of the 1600 F/2000 hour exposure on the tensile and stress rupture properties of the alloy. Tensile strength and ductility at both room temperature and at 1400 F are seriously degraded by the 1600 F exposure as are the stress rupture properties at 1375 F. The room temperature strength and ductility both have been reduced approximately 25% while the 1400 F strength has been lowered nearly the same amount and the ductility has been lowered by over 50%.

IN-738X Alloy. The IN-738X alloy also showed a degrading of the tensile properties as a result of the 1600 F/2000 hour exposure. The resultant reduction of tensile properties in this alloy was nearly the same amount as was found in the Udimet 710 material. While the stress rupture life at 1375 F was reduced about the same as the Udimet 710 alloy (approximately 80%) the ductility was found to increase. All of the specimens tested at this temperature had at least double the elongation of the as heat treated material.

### Electron Vacancy Calculations

The results of the electron vacancy calculations for the analyzed chemistry of the two heats of material used in this investigation and the proposed minimum and maximum specification chemistry for the alloys are shown in Table 8. Calculations were made using the reference method required for this conference as well as by seven other methods which have been used in current work under U.S. Air Force Contract AF 33(615)-5126 by Dr. H. E. Collins at TRW, inc., Cleveland, Ohio. (5,6,7) The two compositions studied proved to have electron vacancy numbers very close to or above the critical values that have been considered safe from a stability viewpoint.

Nv. Actual chemistries for the two alloys calculate very close to the same electron vacancy number, 2.49 for Udimet 710 and 2.47 for IN-738X. Specification minimum and maximum values indicate lower values for IN-738X than for the Udimet 710, but the greater number of elements in the IN-738 X composition causes a greater variation in the

electron vacancy values than for Udimet 710. Based on the results of this work, Nv values as high as 2.49 are indicated to be safe from sigma forming tendencies.

Nv<sub>1</sub> and Nv<sub>2</sub>. These two techniques give erroneous predictions of microstructural instability in the two compositions studied. An Nv of 2.52 has been previously indicated as being the safe upper limit for Nv<sub>1</sub> (8) whereas the Udimet 710 and IN-738X compositions evaluated have values of 2.58 and 2.55. An Nv<sub>2</sub> of 2.53 has been previously indicated to be the safe upper limit with compositions over 2.59 being prone to sigma instability (6) whereas the Udimet 710 and IN-738X compositions studied have values of 2.63 and 2.65, respectively.

 $\underline{\text{Nv}_3}$ . This method has not previously indicated a definite safe value but instability in certain alloys have been found at Nv<sub>3</sub> of 2.23 (PDRL - 163) whereas another alloy, Inco 901 was found to be stable at Nv<sub>3</sub> of 2.28. The compositions tested in this work had 2.33 Nv<sub>3</sub> for Udimet 710 and 2.24 for IN-738X.

Nv<sub>4</sub>. This method accurately predicts the stability found in this study for both alloys, although it indicates slight changes in chemistry in the direction of higher Nv values would cause instability. The Udimet 710 composition has a Nv<sub>4</sub> value of 2.37, the previously established safe upper limit (9). The IN-738X composition has a Nv<sub>4</sub> of 2.33.

Nv5. This method also accurately predicts the stability found in this study and also indicates instability could be expected with only slight chemistry changes. The Udimet 710 and IN-738X compositions, respectively, have Nv5 values of 2.49 and 2.46 whereas the previously established safe upper limit is 2.49 (10).

Nv6. This technique erroneously predicts instability for both compositions. A value of 2.45 for Nv6 has previously been indicated to be the safe upper limit, with sigma forming above 2.51 (6), but the Udimet 710 and IN-738X compositions studied in this work, and found to be stable, have values of 2.53 and 2.58, respectively.

Nv9. Of the techniques used, Nv9 appears to be the least appropriate for the Udimet 710 and IN-738X alloys. The Udimet 710 composition has a Nv9 of 2.94 and the IN-738X, 2.96, whereas the safe upper limit previously established was 2.59, with sigma forming above 2.68.(7)

### Summary of Observations

- 1. Neither IN-738X or Udimet 710 alloy of the composition used in this investigation is prone to form sigma phase even after 2000 hours exposure in the temperature range of 1500 to 1800F.
- 2. Partial solutioning of the gamma prime phase occurs at 1900 F in Udimet 710 and at 1800 F in IN-738X. Complete solutioning of the gamma prime phase occurs only after 1 hour at 2200 F in both alloys.
- 3. The solution temperature of the  $M_{23}C_6$  phase is 2000 F in the Udimet 710 material and 1900 F in the IN-738X alloy.
- 4. M<sub>23</sub>C<sub>6</sub> phase forms at the expense of the MC phase as a function of temperature and time up to 2000 F in the case of the Udimet 710 alloy and up to 1900 F in the case of the IN-738X alloy.
- 5. Exposure at 1600 F for 2000 hours causes a degradation of the tensile and 1375 F stress rupture properties of both alloys.

#### Conclusions

- 1. Although the Udimet 710 and IN-738X compositions studied proved to be stable with respect to sigma formation, their electron vacancy values indicate that only slight composition changes in the direction of higher Nv values could make the alloys strongly prone to sigma formation. A Nv<sub>SS</sub> of 2.49, which is the value of the Udimet 710 composition studied, is shown by this work to be a safe value to resist significant sigma formation.
- 2. The degradation in tensile and 1375 F stress rupture properties caused by the 1600 F/2000 hour exposure is not related to sigma formation.
- 3. The specific microstructural changes which were responsible for the degradation of mechanical properties have not been identified, but should be more thoroughly investigated.

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### References

- (1) Udimet 710, Technical Bulletin No. 1009, Special Metals Corporation, New Hartford, New York.
- (2) IN-738X, "Sulfidation Resistant Alloy" Data Sheet, International Nickel Company, Inc., New York, New York.
- (3) MAR-M Alloy 421, Brochure by Martin Metals, Wheeling, Illinois.
- (4) Private communication from H. B. Canada, Special Metals Corp., 10 June 1968.
- (5) H. E. Collins, "Research on Microstructural Instability of Nickel-Base Superalloys", Interim Engineering Progress Report No. 1, (October 1966), Contract AF 33(615)-5126.
- (6) H. E. Collins and C. S. Kortovich, "Research on Microstructural Instability of Nickel-Base Superalloys", Interim Engineering Progress Report No. 2, (January 1967), Contract AF 33(615)-5126.
- (7) H. E. Collins, "Research on Microstructural Instability of Nickel-Base Superalloys", Interim Engineering Progress Report No. 3, (April 1967), Contract AF 33(615)-5126.
- (8) L. R. Woodyat, C. T. Sims and H. J. Beattie, Jr., "Prediction of Sigma-Type Phase Occurrence from Composition in Austenitic Superalloys", Trans. Met. Soc. AIME, 236, 519 (1966)
- (9) J. W. Slaney, Special Metals, Inc., Oral presentation at the ASTM-ASME Joint Committee on Effect of Temperature on Properties of Materials Gas Turbine Panel Workshop, New York, N.Y., Nov. 29, 1966.
- (10) H. Murphy, General Electric Company, Schenectady, N. Y., Oral Presentation at the ASTM-ASME Joint Committee on Effect of Temperature on Properties of Materials Gas Turbine Panel Workshop, New York, N.Y., Nov. 29, 1966.

Table 1. Chemical Analysis and Specification Range for Udimet 710 and IN-738X Nickel Base Alloys

		Udimet 710	) )		IN-738X	· <u>.</u> .
Alloy		max.	HT 6-5245	min.	max.	HT OM -552
C	0.05	0.10	0.06	0.14	0.20	0.15
Mn		0.10	0.03		0.20	0.10
Si	-	0.20			0.20	0.10
$\mathbf{Cr}_{\mathbf{r}}$	17.50	18.50	17.80	15.50	16.50	15.50
Со	14.50	15.50	15.50	8.00	10.00	8.96
Mo	2.75	3.25	3.17	1.50	2.00	1.68
w	1.25	1.75	1.75	2.30	2.90	2,55
Ti	4.75	5.25	4.97	3.25	3.75	3.75
A1	2.25	2.75	2.48	3.25	3.75	3.70
Zr	<b></b> .	-	<u>.</u> •	0.05	0.15	0.069
Ta		· · · · · · · · · · · · · · · · · · ·	<u> -</u>	1.50	2.00	1.90
Ср		· .		0.60	1.10	1.50
В	0.015	0.025	0.015	0.005	0.015	0.0125
Fe	-	0.50	0.14	- -	0.50	0.14
Cu	<b>-</b>	0.10			0.10	<b>-</b>
S		<b>-</b>	<u>-</u>	-	0.015	
Ni	bal	bal	bal	bal	bal	bal

Table 2. Heat Treatment Schedules for Udimet 710 and IN-738X Nickel Base Alloys

### IN-738X

Solution Treatment - 2050°F - 2 hours - vacuum, back cool with nitrogen to room temperature

Aging Treatment - 1550°F - 24 hours in air - air cool to room temperature

# UDIMET 710

Solution Treatment - 2100°F - 2 hours - vacuum, back cool with nitrogen to room temperature

Aging Treatment - 1950°F - 4 hours - vacuum, back cool with nitrogen to room temperature

Aging Treatment - 1400°F - 16 hours in air - air cool to room temperature

Table 3. Time-Temperature Schedule for Phase Stability Determination of Udimet 710 and IN-738X Nickel Base Alloys

Exposure	Exposure Time, Hours									
Temp, °F		1		10		100		500	1000	2000
						4. 4				
1500			÷					X	X	X
1600								X	X	X
1700		*					- 1.5	X	X	<b>X</b>
1800	: '							X	X	X
1900				$\mathbf{X}^{-}$		X		X	•	
2000		X		X		X				
2100		X								
2200		X								

Table 4. Phases Present in Udimet 710 Alloy After Heat Treatment and Exposure to Elevated Temperature

Specimen	MC	M <sub>23</sub> C <sub>6</sub>	MC:M <sub>23</sub> C <sub>6</sub>	Other
As heat treat	15	3	5:1	Boride + ?
1500/1000 hr	5	10	1:2	Boride or o
1500/2000 hr	2	9	1:4.5	1 1/2 σ ?
1600/1000 hr	3.5	13	1:3.7	Boride or σ ?
1600/2000 hr	1.5	11	1:7.3	1 σ ?
1700/1000 hr	1 .	16	1:16	?
1700/2000 hr	0.25	15.5	1:62	
1800/1000 hr	0.5	17	1:34	?
1800/2000 hr	0.25	19	1:76	
1900/100 hr	5	13.5	1:27	
2000/100 hr	MC onl	y present		
2100/10 hr	MC onl	y present		
2200/10 hr	MC onl	y present (4.31.	Å)	

Table 5. Phases Present in IN-738X Alloy After Heat Treatment and
Exposure to Elevated Temperatures

	X-ray Intensity						
Specimen	MC	M23C6	MC:M23C6	Other			
As heat treat	25	2	12.5:1				
1500/1000 hr	20	7	2.9:1				
1500/2000 hr	15.5	7	2.2:1	1 σ?			
1600/1000 hr	17	9.5	1.8:1				
1600/2000 hr	13	8.5	1.5:1				
1700/1000 hr	17	15.5	1.1:1				
1700/2000 hr	11	14	1:1.3				
1800/1000 hr	12	12	1:1				
1800/2000 hr	17	12	1.4:1				
1900/100 hr	MC onl	y present					
2000/100 hr	MC onl	y present					
2100/10 hr	MC onl	y present					
2200/10 hr	MC onl	y present (4.3	6 Å)				

Table 6. Mechanical Properties of Udimet 710 Alloy as Heat Treated and After Exposure at 1600°F for 2000 Hours

Specimen	Test Temp, °F	UTS ksi	0.2% YS ksi	% Elong in l in.	Reduction of Area %
As heat treat	RT	158.9	130.1	11.2	10.9
1600/2000 hr	RT	129.6	97.6	7.6	9.5
1600/2000 hr	RT	122.0	98.0	7.9	6.9
As heat treat	1400	149.5	112.5	16.4	14.7
1600/2000 hr	1400	112.2	85.5	7.7	6.0
1600/2000 hr	1400	118.6	89.8	8.0	6.9
Specimen	Test C	ondition Loa	d, ksi	Time to Failure, hr	% Elong in l inch
As heat treat	1375		86	137.5	16.7
As heat treat	1375		86	134.8	13.5
1600/2000 hr	1375		86	7.5	10.4
1600/2000 hr	1375		86	22.2	13.5
1600/2000 hr	1375		86	22.1	13.8
As heat treat	1825		16	101.0	29.4
As heat treat	1825		15.3	126.9	14.1
As heat treat	1825		16	145.0	24.7
1600/2000 hr	1825		16	113.3	11.7
1600/2000 hr	1825		16	93.2	15.0
1600/2000 hr	1825		16	104.2	19.5

Table 7. Mechanical Properties of IN-738X Alloy as Heat Treated and After Exposure at 1600°F for 2000 Hours

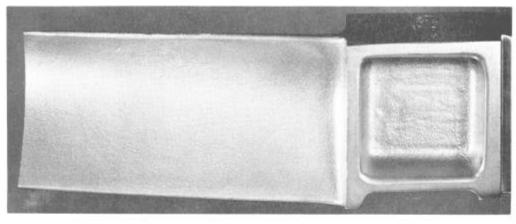
	Test			%	
Specimen	Temp,	UTS ksi	0.2% YS ksi	Elong in l in.	Reduction of Area %
As heat treat	RT	157.3	130.5	7.5	5.3
1600/2000 hr	RT	121.3	104.2	6.0	5.1
1600/2000 hr	RT	105.9	91.5	6.3	4.0
As heat treat	1400	147.2	121.1	18.5	18.3
1600/2000 hr	1400	120.6	100.2	4.0	5.0
1600/2000 hr	1400	123.5	101.8	3.9	4.2
	·	Conditions		Time to	% Elong
Specimen	Temp, °l	<u>Loa</u>	d, ksi	Failure, hr	in 1 inch
As heat treat	Temp, °1		<u>d, ksi</u> 86	Failure, hr 288.3	<u>in 1 inch</u> 3.6
					•
As heat treat	1375		86	288.3	3.6
As heat treat	1375 1375		86 86	288.3	3.6 8.1
As heat treat 1600/2000 hr 1600/2000 hr	1375 1375 1375		86 86 86	288.3 65.7 54.0	3.6 8.1 13.2
As heat treat 1600/2000 hr 1600/2000 hr 1600/2000 hr	1375 1375 1375 1375		86 86 86	288.3 65.7 54.0 52.3	3.6 8.1 13.2 8.7

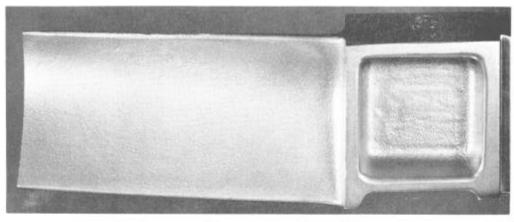
Table 8. Electron Vacancy Calculation Results on Specification Extremes and Analyzed Compositions of Udimet 710 and IN-738X Nickel Base Alloys.

	Udim	et 710		IN-	Previously Proposed		
Nv	Sp		<b>.</b> .			Spec.	
<u>Number</u>	Ht 5245	Min.	Max.	Ht OM 552	Min.	Max.	Safe Value**
Nv <sub>ss</sub>	2.49	2.33	2.66	2.47	2.13	2.60	<b>-</b> -
Nv <sub>1</sub> *	2.58	2.40	2.79	2.55	2.17	2.72	2.52
Nv <sub>2</sub> *	2.63	2.44	2.87	2.65	2.26	2.84	2.53 2.59
Nv <sub>3</sub> *	2.33	2,20	2.48	2.24	1.98	2.37	<b></b>
$^{\mathrm{Nv}_{4}*}$	2.37	2.23	2.54	2.33	2.06	2.48	2.37 2.44
Nv <sub>5</sub> *	2.49	2.33	2.66	2.46	2.12	2.60	2.49
Nv <sub>6</sub> *	2.53	2.36	2.73	2.58	2.21	2.75	2.45 2.51
Nv9	2.94	2.64	3.16	2.96	2.30	3.29	2.59

<sup>\*</sup>Nv numbers shown represent the same methods currently used by Dr. H. E. Collins at TRW, Inc. (5, 6, 7)

<sup>\*\*</sup>Alloys having Nv ≤ this number would not be expected to form sigma phase. Where two values are given alloys having Nv between the numbers may form sigma phase but would not be expected to form sigma phase if the Nv ≤ lower number.











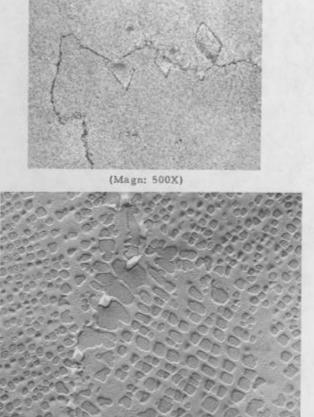


Fig. 2. General microstructure of Udimet 710 in the fully heat treated condition.

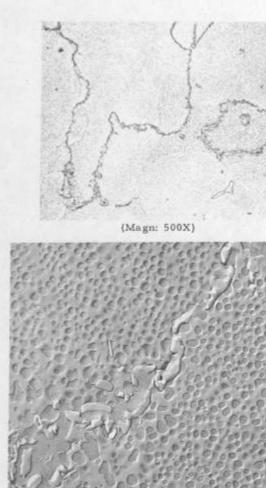


Fig. 3. General microstructure of Udimet 710 in the plus 1500F - 1000 hour - air cool condition.

(Magn: 5000X)

(Magn: 500X)

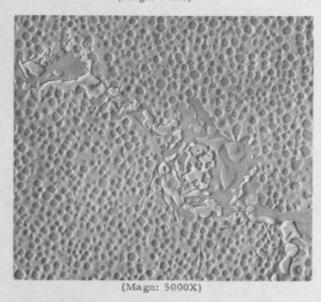
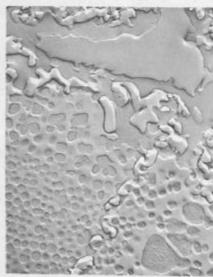


Fig. 4. General microstructure of Udimet 710 in the fully heat treated plus 1500F - 2000 hour - air cool condition.



(Magn: 500X)



(Magn: 5000X)

Fig. 5. General microstructure of Udimet 71 plus 1600F - 1000 hour - air cool cor

Fig. 6. General microstructure of Udimet 710 in the fully heat treated plus  $1600\,\mathrm{F}$  -  $2000\,\mathrm{hour}$  - air cool condition.

Fig. 7. General microstructure of Udimet plus 1700F - 1000 hour - air cool c

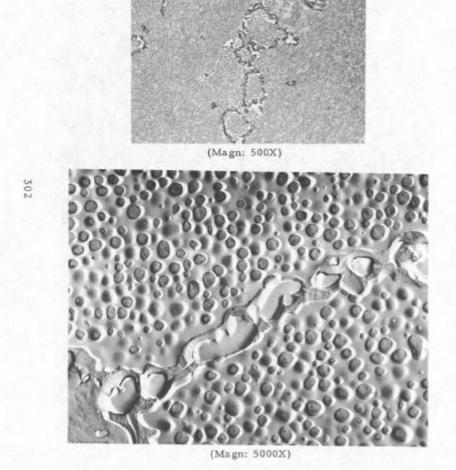
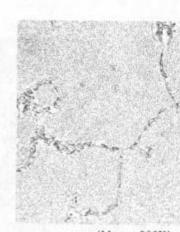
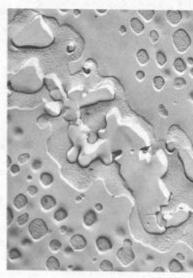


Fig. 8. General microstructure of Udimet 710 in the fully heat treated plus 1700F - 2000 hour - air cool condition.

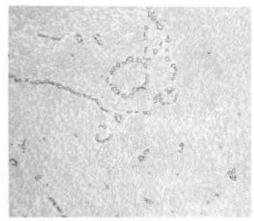


(Magn: 500X)



(Magn: 5000X)

Fig. 9. General microstructure of Udime plus 1800 F - 1000 hour - air coo



(Magn: 500X)

303

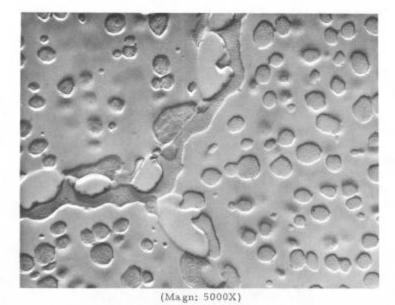
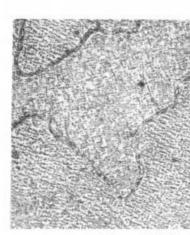


Fig. 10. General microstructure of Udimet 710 in the fully heat treated plus 1800F - 2000 hour - air cool condition.



(Magn: 500X)

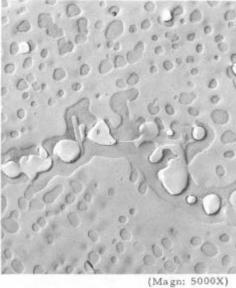
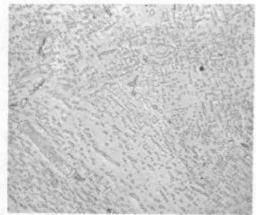
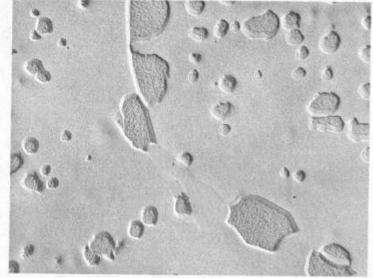


Fig. 11. General microstructure of Udimet treated plus 1900F - 500 hour - si

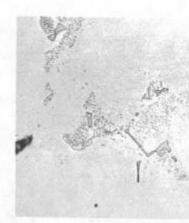


(Magn: 500X)

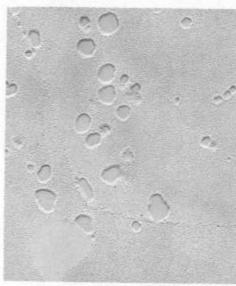


(Magn: 5000X)

Fig. 12. General microstructure of Udimet 710 in the fully heat treated plus 2000 F - 500 hour - air cool condition.



(Magn: 500X)

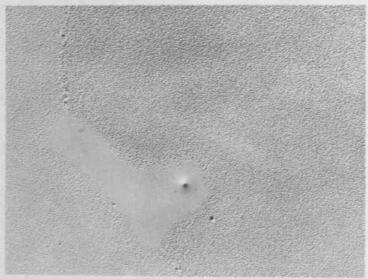


(Magn: 5000X)

Fig. 13. General microstructure of Udin treated plus 2100F - 10 hour -

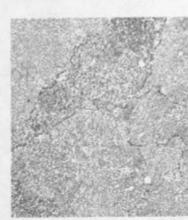


(Magn: 500X)

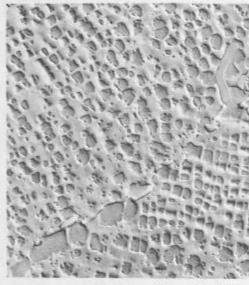


(Magn: 5000X)

Fig. 14. General microstructure of Udimet 710 in the fully heat treated plus 2200 F - 10 hour - air cool condition.



(Magn; 500X)



(Magn: 5000X)

Fig. 15. General microstructure of IN-7 treated condition.

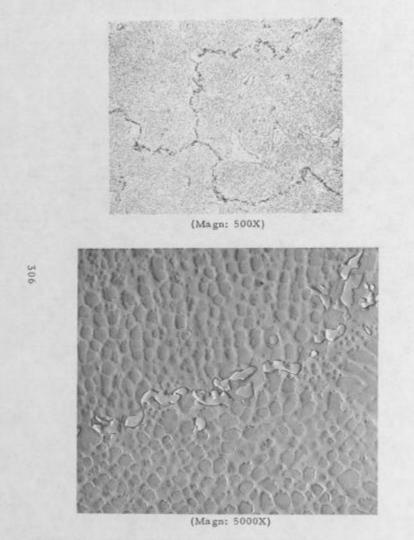
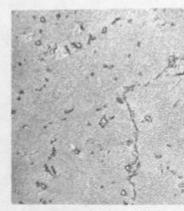


Fig. 16. General microstructure of IN-738X in the fully heat treated plus 1500 F - 1000 hour - air cool condition.



(Magn: 500X)

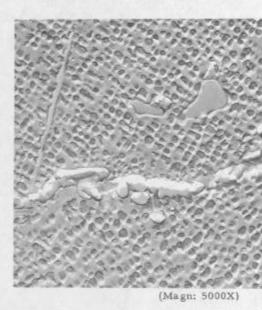


Fig. 17. General microstructure of IN-738X plus 1500F - 2000 hour - air cool c

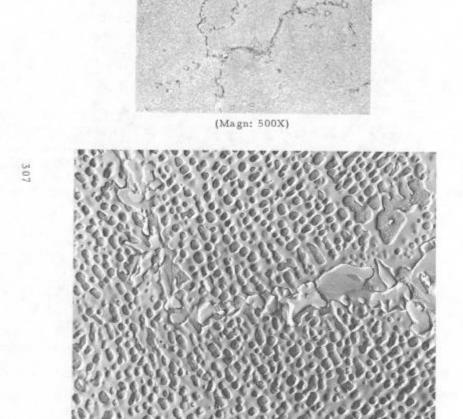
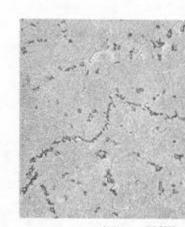


Fig. 18. General microstructure of IN-738X in the fully heat treated plus 1600F - 1000 hour - air cool condition.



(Magn: 500X)

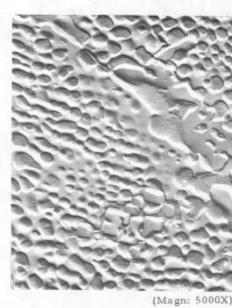


Fig. 19. General microstructure of IN-7 plus 1600 F - 2000 hour - air co

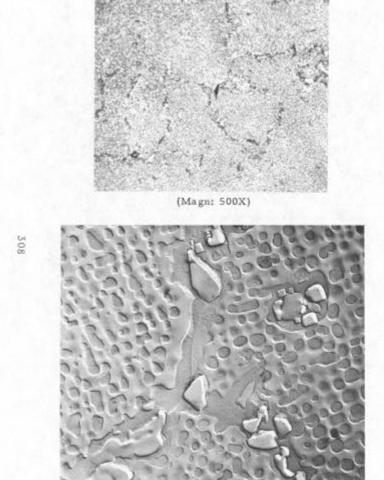
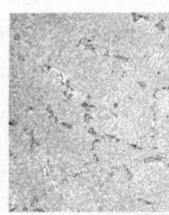
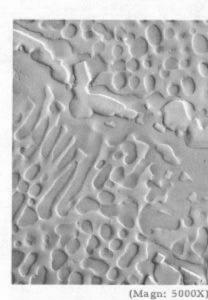


Fig. 20. General microstructure of IN-738X in the fully heat treated plus 1700F - 1000 hour - air cool condition.

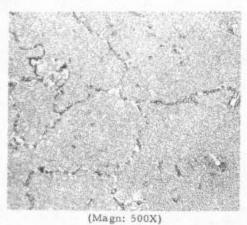


(Magn: 500X)



- Value Takey ----

Fig. 21. General microstructure of IN-7 plus 1700 F - 2000 hour - air co



309

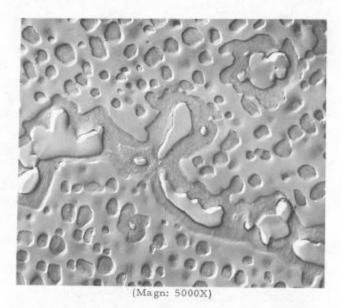
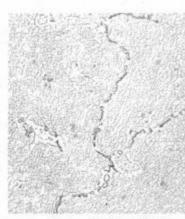
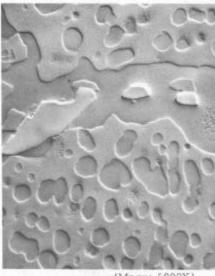


Fig. 22. General microstructure of IN-738X in the fully heat treated plus 1800F - 1000 hour - air cool condition.



(Magn: 500X)



(Magn: 5000X)

Fig. 23. General microstructure of IN-738X plus 1800 F - 2000 hour - air cool o

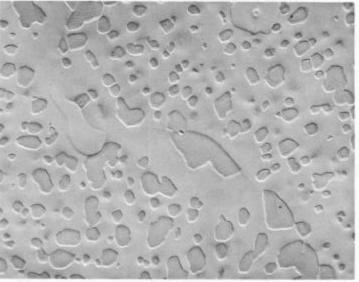
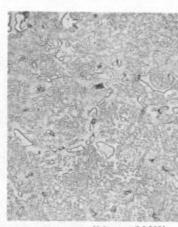


Fig. 24. General microstructure of IN-738X in the fully heat treated plus 1900 F - 500 hour - air cool condition.



(Magn: 500X)

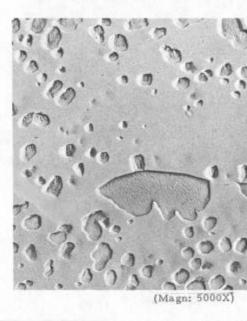
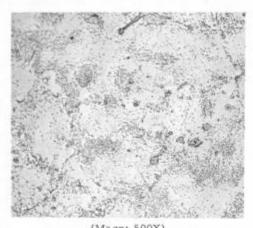


Fig. 25. General microstructure of IN-738 plus 2000F = 500 hour - air cool of



(Magn: 500X)

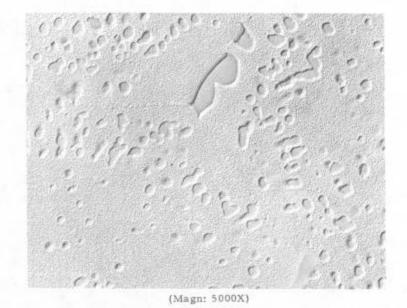
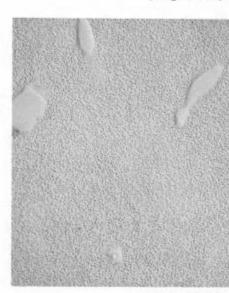


Fig. 26. General microstructure of IN-738X in the fully heat treated plus 2100 F - 10 hour - air cool condition.



(Magn: 500X)



(Magn: 5000X)

Fig. 27. General microstructure of IN-7 plus 2200F - 10 hour - air cool