EFFECT OF CHEMISTRY VARIATIONS ON THE STRUCTURAL STABILITY OF ALLOY 718

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Abstract

The effects of thermo-mechanical processing and heat treatment on the properties and phase stability of Alloy 718 have been investigated in detail. Although much work has been conducted on the microstructural characteristics of this alloy, variations in the appearance of standard microstructures are continually being reported. Because of the great interest in this alloy by producers and users, the current program was established to determine the effects of variations in alloying elements on the microstructure and mechanical properties of Alloy 718.

The procedure involved melting a series of experimental heats with wide variations in columbium, titanium, aluminum, and silicon. The program was statistically designed to determine the effect of these various elements singularly and in combination on the structure and properties of this alloy. This study was conducted on samples in two as-heat treated conditions and after exposure at 1200°F for an extended period of time.

The microstructural characteristics of the various compositions have been evaluated by light and electron microscopy. Secondary phases present in these compositions have been identified by X-ray diffraction analysis of electrolytically extracted residues. An attempt has been made to correlate microstructure, mechanical properties, and chemistry for the experimental heats studied in this investigation.

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Introduction

For the past seven years a considerable amount of time and effort has been expended by both suppliers and users in the study of Alloy 718. The interest in this material stems not only from the interesting metallurgy associated with the alloy, but with the large usage of the material in numerous applications. In fact, many people believe that the projected use of this alloy will increase significantly over the next several years. The large use of Alloy 718 is the result of its performance in respect to high strength, good ductility, good repair weldability, and from its versatility of being supplied in various product forms.

The current study was initiated to determine the effect of chemistry variations on the mechanical property and phase stability characteristics of Alloy 718. The elements studied included columbium, titanium, aluminum and silicon. There is nothing unique involving studies of the first three elements, since a number of previous investigators (1-3) have looked at variations of this type in considerable detail. The primary feature of this work is the variation in silicon content in combination with variations in the other three elements both within and outside most specification limits for Alloy 718.

The modified compositions of Alloy 718 in this study were designed to be statistically evaluated by multiple regression analyses. Mechanical properties determined included room temperature tensile and impact tests on material receiving two different heat treatments. Additional material was exposed at a representative service temperature of 1200°F for an extended period of time and also subjected to room temperature tensile tests. Additional data on the structure were obtained on various samples using light and electron microscopy in conjunction with X-ray diffraction analyses of electrolytically extracted residues.

The tensile and impact test data on heat treated and heat treated plus exposed material were correlated with variations in the chemical analyses. Correlation of chemistry with phases present in heat treated and in heat treated plus exposed samples was also attempted. Finally, correlations between microstructure and measured properties were attempted for the experimental heats. The electron vacancy concept $(N_{\rm V})$ was applied in an attempt to establish a possible relationship among chemistry, properties and the formation of various intermetallic phases.

Procedure

Material

The material used in this study was processed from a series of 17 laboratory heats. Each heat was vacuum induction melted, cast into 2-1/4 inch diameter electrodes and subsequently vacuum arc remelted into 3-9/16 inch round ingots. Virgin electrolytic materials were used for all heats and a uniform melt practice maintained for all heats.

Table I contains the nominal range of the various elements in Alloy 718 as required by Aeronautical Materials Specifications 5662B and 5664A. Also included in this table are the chemical analyses of the experimental compositions melted for this investigation. The data show variations in the columbium, titanium, aluminum and silicon contents of these heats. The elements were varied to have at least one heat within the specified AMS chemistry range. The remainder of the heats exhibited chemistry variations on the low and high side of the nominal Alloy 718 chemistry range.

Fabrication

All of the experimental vacuum arc remelted ingots were processed in a similar manner. The ingots were charged into a furnace at 2025°F, soaked for two hours and rolled to an intermediate size of two inch round cornered square bar. A 1700°F minimum working temperature was maintained during this initial processing and the hot working characteristics of all heats were good. After being surface ground to remove small cracks, the two inch round cornered square bars were recharged into a furnace set at 2025°F, soaked for two hours, and then rolled to 15/16 inch round cornered square bars. Again hot workability was excellent on all heats. Following another light grinding the bars were hot rolled to a final size of 1/2 inch using the same hot working temperature.

Testing

Three groups of specimens were machined from each of the experimental Alloy 718 compositions. One group was solution treated at 1750°F for one hour and air cooled. These specimens were than aged at 1325°F for eight hours, furnace cooled to 1150°F, held an additional eight hours, and then air cooled. The second group of specimens was solution treated at 1950°F for one hour, air cooled and given the same aging treatment. Both groups of specimens were then room temperature tensile and impact tested.

The third group of samples was solution treated at 1750°F for one hour, air cooled and then aged at 1325°F for eight hours, furnace cooled to 1150°F, held eight hours and then air cooled. These samples were then given an additional exposure treatment at 1200°F for 1500 hours to simulate use in service. Upon completion of this exposure the samples were tensile tested.

All tensile test specimens had a 0.160 inch gage diameter and a 1.0 inch gage length. Tests were conducted at a strain rate of 0.005 inch per inch per minute to the 0.2 percent yield point and at 0.05 inch per inch per minute beyond the yield. Recorded data include ultimate strength, 0.2 percent yield strength, elongation and reduction of area. Duplicate impact tests were conducted on standard Charpy V-Notch specimens with the strength being recorded in foot-pounds.

X-Ray Diffraction Analyses

Specimens from each of the experimental heats in the heat treated condition (1750°F solution treatment plus age) and the heat treat-

ed plus exposed condition (1500 hours at 1200°F) were subjected to X-ray diffraction analysis. The secondary phases in these alloys were electrolytically extracted and the residues analyzed by standard X-ray diffraction techniques.

Because of the various types of phases present in Alloy 718, two extraction solutions were employed. One solution consisted of 10 percent hydrochloric acid in water and the other of 10 percent hydrochloric acid in methanol. For the 10 percent acid in water extractions, the specimens were turned and cleaned in the extraction solution for 10 minutes at three volts and 0.5 amps per square inch. The cleaned samples were then washed with alcohol, placed into a fresh solution and the extraction conducted for 20 minutes at three volts and 0.5 amps per square inch. The extracted residue was washed in methanol a number of times to remove any traces of the acid. This solution was employed to obtain the fine matrix precipitates (i.e., gamma prime and Ni₃Cb). The extracted residue was dried, attached to a glass slide, held in place by parlodion and scanned using the X-ray diffractometer.

For the 10 percent hydrochloric acid in methanol solution the specimens were lathe turned and cleaned in the extraction solution for 30 minutes, at two volts and 0.1 amps per square inch. After being cleaned in methanol, the samples were placed into a fresh solution and the extraction conducted at 0.1 amps per square inch and two volts for four hours. The residue was washed, allowed to settle for two hours, decanted, and washed several times in methanol to remove all traces of the acid. This solution was primarily used to extract the larger secondary precipitates and/or carbides (i.e., Laves, CbC, and M₆C). The dried residues were fastened to a glass slide with parlodion and subjected to X-ray diffraction analysis.

Microstructural Studies

Optical microscopy studies were conducted on all heats for grain size determinations and to observe gross variations in the structure associated with carbide and intermetallic phase formation. A number of samples were also examined by electron microscopy to study the fine matrix and grain boundary precipitates. Sample preparation for the electron microscope involved normal grinding through 600 grit paper and wheel polishing through Linde-A alumina abrasive. The specimens were then electro-polished for five seconds using a 10 percent hydrochloric acid in methanol solution. Light chemical etching in hydrochloric acid, methanol and H₂O₂ was required to complete the sample preparation and leave the secondary phases in relief. The specimens were replicated with parlodion and shadowed with chromium for contrast.

Multiple Regression Analyses

The experimental compositions melted for this program were designed to be analyzed by multiple regression techniques. These analyses were conducted by a computer programmed to study six variations on any given run. This program, therefore, allowed any combination of six linear, quadratic, cubic or interaction effects as-

sociated with chemistry changes to be analyzed. The six most significant composition variables were determined on the basis of the greatest amount of variation in a given property accounted for by the analyses. This procedure enabled the formulation of regression equations for all of the properties and phases determined in this study.

Results

Tensile Properties

The results of the room temperature tensile tests conducted on samples solution treated at 1750°F and 1950°F are listed in Tables II and III respectively. Although two different solution treating temperatures were used, all samples received a common aging treatment of 1325°F for eight hours, furnace cooling to 1150°F, holding for eight hours and then air cooling. The tensile and 0.2 percent yield strengths obtained on each of the experimental heats are illustrated graphically in Figure 1. Also included in Figure 1 are the minimum requirements for tensile and yield strength required by AMS specifications 5662B and 5664A. The data are plotted in a descending order of yield strength for clarity of interpretation and show a tensile strength range of 167,000 to 220,000 psi for materials solution treated at 1750°F. The same practice was followed in succeeding figures; therefore heat numbers are not necessarily in the same order. The range of tensile strengths for the material solution treated at 1950°F are slightly lower than the strengths obtained on materials solution treated at 1750°F and range from 153,000 to 207,000 psi. Most of the experimental heats meet the minimum tensile strength requirements except for a few, all of which had columbium contents lower than the nominal AMS specification range for this element.

The range of yield strengths for material solution treated at 1750°F varied considerably and ranged from a low of 93,000 to a high of 198,000 psi. Yield strength values for specimens solution treated at 1950°F were generally lower and varied from a low of 98,000 psi to a high of 179,000 psi. More of the experimental heats failed to meet the minimum yield strength than the minimum tensile strength apparently indicating more sensitivity to chemistry variations. Most of the low yield strengths were associated with heats with low columbium contents.

The variation in tensile ductility as measured by both elongation and reduction of area is illustrated in Figure 2. Also plotted in Figure 2 are the minimum ductility requirements specified by AMS specification 5662B and 5664A. The experimental heats exhibited a wide variation in both elongation and reduction of area. Elongation values ranged from a high near 35 percent to a low near eight percent while reduction of area values ranged from a high of 60 percent to a low of eight percent. Most of the experimental heats, however, met the minimum ductility requirements of the AMS specification. The tensile ductility data also show that material solution treated at 1950°F exhibit higher ductilities for each of the experimental heats. As would be expected, those heats with lowest ductilities exhibited the highest strength values.

Tensile Properties After Exposure

The results of tensile tests conducted on material fully heat treated (1750°F solution temperature plus age) and exposed at 1200°F for 1500 hours are listed in Table IV. These data are illustrated in Figure 3 for each of the experimental heats and compared with the tensile and yield strength values obtained on heat treated specimens not exposed at 1200°F. All of the heats (with few exceptions) show an increase in both tensile and yield strength after exposure at 1200°F for 1500 hours. The effect of this exposure on the ductility is illustrated in Figure 4. There appears to be a slight decrease in ductility associated with the increased strength of the exposed samples, but in most instances, the variation between heat treated and heat treated plus exposed ductility values do not differ greatly.

Impact Properties

The results of impact tests conducted on specimens solution treated at 1750°F and 1950°F are listed in Table V. These data, illustrated in Figure 5, show a wide variation of impact strength. Values range from a high of 75 foot-pounds to a low of approximately nine foot-pounds for material solution treated at 1950°F. Samples solution treated at 1750°F have slightly lower values and range from approximately 70 foot-pounds to a low of seven foot-pounds. Tensile elongation values were also plotted in Figure 5 for purposes of comparison with the impact strength. The impact strength and tensile elongation follow the same trends. Again, the principle element involved in impact strength appeared to be columbium since most of the heats with extremely high impact strengths had low columbium contents.

X-Ray Diffraction Analyses

X-ray diffraction analyses were conducted on residues extracted from each of the experimental heats in the fully heat treated (1750°F solution temperature) and the fully heat treated plus exposed conditions. The extracted residues were obtained using two different solutions. The initial solution was 10 percent hydrochloric acid in methanol and the second solution was 10 percent hydrochloric acid in water. As described in the procedure, the first was used primarily to extract large second phase particles such as the various carbides and Laves phase. The second solution was used to obtain the finer intermetallic precipitates such as gamma prime and Ni₃Cb.

The results of the X-ray diffraction analyses on extracted residues from each of the two solutions are listed in Table VI for samples solution treated at 1750°F followed by the standard aging treatment. Six phases have been identified as major constituents in these experimental Alloy 718 heats. The phases identified include CbC, Ni3Cb, gamma prime, NixCb, Laves and M6C (high lattice parameter).

No chemical analyses were conducted on the extracted residues, but it is assumed that the CbC carbide is primarily a columbium-rich carbide with some solubility for titanium and possibly molybdenum and chromium. The M₆C carbide phase normally contains molybdenum as the major metallic element but can also contain some nickel and cobalt. The lattice parameter of this phase, as found in these alloys, is larger than the usual M₆C phase found in most nickel base superalloys.

indicating the possible presence of columbium in this phase.

The Laves phase found in these heats is usually thought to be Fe₂Ti, Fe₂Mo, or Ni₂Cb, although other elements such as silicon and chromium can also be in solution in this phase. (4) The gamma prime found in these analyses is primarily Ni₃Al with some solubility for titanium and columbium. The Ni_xCb has previously been identified by Radavich(5) and is believed to be a transition phase from the face centered cubic gamma prime to orthorhombic Ni₃Cb. Ni_xCb has body centered tetragonal structure. The Ni₃Cb phase has an orthorhombic crystal structure and is composed almost solely of nickel and columbium.

The relative amount and relationship of Laves, CbC, and M_6C phases in the heat treated experimental Alloy 718 heats are illustrated in Figure 6. The heats were sequenced in order of decreasing amounts of Laves phase. With heats arranged in the same order, M_6C appears to peak at the center with low values at either end of the sequence. The amount of CbC, on the other hand, is relatively constant throughout most of the heats except for an increase in those heats that have relatively low concentrations of both Laves and M_6C . These data appear to indicate that columbium is contained in both Laves and M_6C .

Figure 7 illustrates relative amounts and relationship among gamma prime, Ni₃Cb, and Ni_xCb for each of the fully heat treated experimental Alloy 718 heats. There appears to be little correlation between the gamma prime and the Ni₃Cb phases. However, the data do show that the appearance of the body centered tetragonal phase Ni_xCb coincides with the absence of any observable gamma prime in the material. This would tend to confirm that Ni_xCb is a transition phase.

X-ray data obtained on extracted residues of samples fully heat treated and exposed at $1200^{\circ}F$ for 1500 hours are listed in Table VII. The same six phases appear in the exposed material as were found in the as heat treated samples. The exposed samples contained CbC, M₆C, and Laves in amounts approximately equal to the amounts observed in samples fully heat treated with no exposure. These data indicate that exposure at $1200^{\circ}F$ has little or no effect on the carbide phases or on the Laves phase.

The only phases which appear to have been affected by the 1500 hour exposure at 1200°F are those which normally precipitate in this temperature region. Gamma prime was found in all but three of the experimental heats after exposure at 1200°F. This represents no change from the data obtained on heat treated samples. In those heats containing no gamma prime, Ni₃Cb and/or Ni_xCb appear in abundance and are the precipitates responsible for the strengthening. The change in the precipitating phases upon exposure at 1200°F are illustrated in Figure 8 which shows the relative intensities of the gamma prime, Ni₃Cb and Ni_xCb. Three heats, which did not exhibit Ni₃Cb in the heat treated condition, show the appearance of this phase after the 1500 hour exposure, indicating that the precipitation took place during the exposure treatment.

Microstructure.

Specimens from each of the experimental heats solution treated at 1750°F and 1950°F and then given the standard aging treatment were examined by optical and electron microscopy. Typical structures observed by optical microscopy are illustrated in Figure 9. Grain size was relatively fine for all specimens, averaging ASTM 7 to 8. The grain size range for all of the heats varied between ASTM 5.5 and 10. Generally, samples solution treated at 1950°F had slightly coarser grain sizes than the material solution treated at 1750°F. However, there appeared to be no relationship between the measured grain size and the mechanical properties.

Figure 10 illustrates the structure, as observed on the electron microscope, of a specimen fully heat treated using a solution temperature of 1750°F. The long, "needle-like" precipitates in this structure are orthorhombic Ni₃Cb particles. The angular grain boundary precipitates are CbC carbides. The less angular, almost circular particles observed in this structure are believed to be Laves phase, This was confirmed by microprobe analysis of several particles which indicated that they were rich in columbium, molybdenum, and silicon but contained no carbon. The presence of these elements in this Laves phase has been verified by other investigators. (6) In the matrix, a fine, almost unresolvable precipitate appears, which based on X-ray data is believed to be body centered tetragonal Ni_xCb. This structure also shows Ni₃Cb needles emanating from some of the other precipitates. Since both the CbC carbides and the Laves phase are rich in columbium, these phases are apparently feeding the Ni₃Cb phase with the columbium required for its growth.

Figures 11, 12, and 13 show the structure of fully heat treated plus exposed samples as observed on the electron microscope. Figure 11 shows a very fine "needle-like" precipitate appearing throughout the matrix which has been identified as Ni₃Cb. The appearance of this phase resulted after the 1500 hour exposure at 1200°F. Figure 12 shows not only the fine "needle-like" Ni₃Cb phase precipitated in the matrix, but in addition, the large Ni₃Cb particles which were present in the heat treated condition. Figure 13 shows a fine grain boundary precipitate in this particular heat which is believed to be the M₆C particles. The actual composition of this phase could not be determined since these particles are too small to be scanned on the microprobe. A very fine gamma prime precipitate can also be seen throughout the matrix of this particular heat.

Discussion

Multiple regression analyses were conducted on all mechanical properties and X-ray data generated by this investigation. A maximum of six independent variables could be employed at any one time. Thus, the results show the six most significant variables obtained for each property or X-ray result analyzed. Both the regression equation and the amount of variation of any given property accounted for by the equation was determined. Those equations producing values above 80 percent are highly significant, while those equations with percentages below 80 percent are less significant but the trends can be seen and

are often just as important. The results of the multiple regression analyses are discussed in the following sections.

Tensile Properties

Heat Treated. The multiple regression equations for the tensile properties are listed in Table VIII. The amount of variation in properties accounted for by these equations ranges from 72 to 93 percent. By inserting values for the variable elements into these equations, the effect of composition on tensile properties can be determined. Figures 14 and 15 show the trends of the tensile data for the specimens solution treated at 1950°F.

Figure 14 shows that columbium is the most important element controlling the strength characteristics of Alloy 718. Increasing the columbium content strongly increases the tensile strength of material solution treated at 1950°F. A similar effect can be seen for the 1750°F solution treatment. The effect of increasing titanium shows a moderate increase in tensile strength while the effect of aluminum is variable, decreasing at first, reaching a minimum between 0.7 and 0.85 percent aluminum and increasing thereafter. Increased silicon has no significant effect on tensile strength on material solution treated at 1750°F and only slightly increases tensile strength of material solution treated at 1950°F.

The effect of chemistry variations on tensile ductility is illustrated in Figure 15. Increasing the columbium and silicon content results in moderate decreases in both elongation and reduction of area. Increasing titanium content has no effect on ductility of material solution treated at 1750°F and shows a very slight decrease in ductility on material solution treated at 1950°F. Aluminum content has no significant effect on tensile elongation and reduction of area at either solution treating temperature.

Heat Treated Plus Exposed. The effect of exposure at 1200°F for 1500 hours and chemistry variations on the tensile properties of Alloy 718 is illustrated in Figure 16. Increasing columbium content shows strongs increases in tensile and yield strengths, whereas increasing titanium content shows moderate increases in the strength. While not plotted, increased aluminum did cause significant decreases in the tensile strength after exposure. Silicon content had no apparent effect on strength in the exposed condition.

Increasing columbium and silicon content showed moderate decreases in the elongation and reduction of area values of exposed samples. Titanium and aluminum showed no effect on the elongation but increasing amounts of these elements did cause a slight decrease in the reduction of area values.

Impact Properties

The regression equations determined from the impact data are listed in Table IX. The effect of increasing columbium and silicon contents on the impact properties of material solution treated at 1750 and 1950°F is illustrated in Figure 17. Increased amounts of both columbium and silicon cause a significant decrease in the impact strength at both solution treatment temperatures. Increased amounts

of aluminum and titanium show a slight decrease of impact strength on material solution treated at 1750°F but had no significant effect on material solution treated at 1950°F.

X-Ray Diffraction Analysis

The results of the multiple regression analyses on the X-ray data are listed in Table X. Six phases were determined from this series of Alloy 718 heats. The phases included CbC, M₆C, Laves, gamma prime, Ni₃Cb, and Ni_xCb. The amount of CbC carbide was significantly affected by three of the four elements varied as shown in Figure 18. Increased amounts of columbium and silicon strongly decreased the tendency for the formation of CbC carbide in heat treated and in heat treated plus exposed material. Aluminum and titanium had no effect on the CbC carbide in heat treated material. Similarly, titanium had no effect on CbC carbide in material exposed at 1200°F. However, increased amounts of aluminum revealed a significant increase in the amount of CbC carbide for material exposed at 1200°F for 1500 hours.

Increased amounts of columbium, aluminum and titanium had no effect on the formation of the M₆C carbide in either heat treated or heat treated plus exposed material. Silicon, on the other hand, exerted a very significant effect with increased amounts of this element resulting in increased amounts of M₆C carbide.

The effect of chemistry variations on the formation of Laves phase is illustrated in Figure 19. These data show an increased tendency for Laves phase formation with increased concentrations of columbium, aluminum and silicon in Alloy 718. This effect was observed on heat treated material as well as material heat treated and exposed at 1200°F for 1500 hours. Titanium had no effect on Laves phase in either the heat treated or the heat treated plus exposed condition.

The effect of chemistry variations on the presence and amount of gamma prime is illustrated in Figure 20. Increasing amounts of columbium caused a slight decrease in the amount of gamma prime in heat treated and heat treated plus exposed material. Increased aluminum content resulted in strong increases in the amount of gamma prime substantiating the importance of aluminum in this phase. The effect of increasing titanium was variable but showed only a slight general increase in the amount of gamma prime regardless of condition. Silicon had no effect on the presence of gamma prime in either the heat treated or the heat treated plus exposed material.

All chemistry variations were observed to have an effect on the presence of Ni₃Cb in both heat treated and heat treated plus exposed material. The effects of exposed material are illustrated in Figure 21. Increased columbium resulted in increased amounts of Ni₃Cb, whereas increased amounts of aluminum caused a slight decrease in the amount of this phase. Increased amounts of silicon caused slight to moderate decreases in the Ni₃Cb phase. Although not illustrated, similar effects occurred in as-heat treated material. The effect of titanium on Ni₃Cb formation was variable. Increased amounts of titanium produced a strong increase in the presence of this phase in heat treated material, but after exposure at 1200°F, increased titanium resulted in a decreased tendency for the formation of the Ni₃Cb phase.

Electron Vacancy Number $(N_{V_{SS}})$

The electron vacancy number (N_{vss}) for each of the experimental heats studied in this investigation are listed in Table I along with the chemical composition of the individual heats. The calculation of the electron vacancy number was based on the method requested by the Symposium committee so as to have common basis of comparison. The assumptions made in the calculation of this electron vacancy number are as follows:

1. Half the carbon forms MC (i.e., CbC and TiC)

2. Half the carbon forms M23C6[i.e., (Cr21Mo2)C6]

3. M₃B₂ boride is formed [i.e., (Mo_{0.5}, Ti_{0.15}, Cr_{0.025}, Ni_{0.10})₃B₂] 4. Gamma prime is of the following composition:

Gamma prime is of the following composition [Ni₃(Al, Ti, Cb, 0.3 Cr)]

5. The electron vacancy number is calculated using the following equation:

 $N_{V_{SS}} = 0.66 \text{ Ni} + 1.71 \text{ Co} + 2.66 \text{ Fe} + 3.66 \text{ Mn} + 4.66 \text{ (Cr+Mo+W)} + 5.66 \text{ V} + 6.66 \text{ Si}$

Even though the calculation of the electron vacancy number is based on the alloy chemistry, a multiple regression analysis was conducted to see the effect of the variable elements. The resultant regression equations are listed in Table X and show that the percent of variation accounted for in the electron vacancy number was in excess of 98 percent. All of the variable elements are included in the equation which describes the electron vacancy number and in every case there is less than one percent difference between the electron vacancy numbers calculated by the reference method and those calculated by the equation in the table. Each of the elements studied in this investigation produced an increase in the electron vacancy number with an increase in alloying content.

The electron vacancy number (N_{VSS}) had no mathematical correlation with any of the phases formed during heat treatment or after exposure at 1200°F for 1500 hours. Similarly, there was no observed correlation of electron vacancy number with any of the measured properties. This lack of correlation may be due to incorrect assumptions used in calculating the N_{VSS} number. For example, no $M_{23}C_6$ or M_3B_2 phases were observed in the material even though they were included in the calculations.

An attempt was made to establish another method of calculating the electron vacancy number. This could not be accomplished effectively because of lack of knowledge of meaningful chemical compositions of the microconstituents. This work has shown columbium and silicon are intimately involved in the formation of Laves, CbC and M_6 C. The presence of Laves in most of the heats studied would indicate that the compositions are above the critical electron vacancy regardless of the method of determination.

Conclusions

The following conclusions were reached based upon data obtained from a series of experimental Alloy 718 heats with chemistry variations previously noted:

Ni₃Cb, and gamma prime. Increased amounts of M₆C carbide appear to decrease the strength capabilities of material solution treated at 1750°F. This effect is not apparent on samples solution treated at 1950°F.

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- 1. Of the variable elements studied, the effect of columbium was the most significant. Increasing amounts of columbium resulted in significantly higher strength for heat treated and heat treated plus exposed material. The increase in strength was reflected in lower tensile ductility and lower impact strength. Increasing amounts of columbium also resulted in an increased tendency for formation of Ni₃Cb and Laves phase. Low columbium contents tended to result in the formation of more gamma prime and CbC carbide.
- 2. The effect of aluminum variables was much less significant and considerably more variable than columbium. Increased aluminum content increased the tendency for the formation of Laves phase and exhibited either no effect on a variable effect on the tensile and impact strength as well as the M₆C carbide phase for heat treated and heat treated plus exposed material. In some instances, low aluminum content favored improved ductility and also increased the tendency for the formation of Ni₃Cb.
- 3. The effect of titanium was important in respect to properties but to a lesser extent than columbium. High titanium content generally produced higher strength and increased amounts of gamma prime and Ni₃Cb in heat treated material. In certain instances, low titanium content favored slightly improved ductility. Titanium had no significant effect on the remaining properties and phases analyzed.
- 4. The effect of silicon followed a trend similar to that of columbium with a few exceptions. Increased amounts of silicon either increased the strength slightly or had no significant effect on this property. High silicon promoted the formation of Laves phase as did columbium but in addition also promoted the formation of the M6C carbide. Low levels of silicon resulted in improved tensile ductility and impact strength as well as favoring the formation of Ni₃Cb and the CbC carbide.
- 5. In respect to the other variables studied in this investigation, the results show that higher solution treating temperatures produce lower room temperature tensile and yield strength and higher tensile ductility and impact strength. Exposure at 1200°F for 1500 hours produced only slight effects on the tensile properties, usually resulting in slightly higher strength and lower ductility. Exposure at 1200°F had little or no effect on the M6C carbide or Laves phase, but exposure at this temperature did result in increased precipitation of gamma prime, Ni3Cb and in certain cases the CbC carbide.
- 6. The overall results and relationship among the various factors studied showed no correlation between electron vacancy number ($N_{\rm VSS}$) and either the properties or phases determined by X-ray diffraction analysis. All of the variable elements studied which included columbium, aluminum, titanium and silicon increased the calculated electron vacancy number. The correlation between phases and properties is somewhat indeterminant, but high strength and reduced ductility are associated with increased amounts of Laves phase,

TABLE I

Chemical Compositions of the Variations in Alloy 718

		***		-	Chemical Composition (weight percent) for:										
]	Heat Number	<u>C</u>	Mn	S	<u> </u>	Cr	<u>Ni</u>	Fe	Мо	Co	В	Ti	Al	C	
	Specification AMS	0.08	0.35	0.15 max	0.015	17.00	50.00	Bal.	2.80	1.00	0.006	0.65	0.20		
	5664 A	mex	max (max	max	21.00	55.00		3.30	max	max	1.15	0.80	5•!	
]	5662 B Experimental Aim	0.057	0.11	0.009	0.005	17.73	53.00	Bal.	3.07	0.30	0.0047	VARIA	ABLE EI	LEMEI	
]	cc 1539	0.062	0.14	0.007	0.004	17.48	52.94	18.09	3.01	0.23	0,0084	0,91	0.68	5.1	
	CC 1540	0.063	0.11	0.007	0.005	17.34	53.34	18.46	3.00	0.26	0.0083	0.94	0.67	5.1	
	CC 1541	0.064	0.11	0,011	0.011	17.60	52.96	18.87	2.96	0.36	0.0075	0.37	0.74	5.3	
	CC 1542	0.060	0.11	0.009	0.004	17.42	53.21	18.03	2.99	0.36	0.0090	0.97	0.75	5.1	
ŀ	C 1543	0.058	0.11	0.014	0.005	17.40	54.20	14.94	3.02	0.36	0.0099	1.37	1.12	6.1	
	C 1544	0.071	0.11	0.009	0.001	17.80	53.27	18.06	2.94	0.39	0.0055	0.52	0.31	6.0	
	C 1545	0.062	0.14	0.009	0.003	17.89	52.75	19.20	2.93	0.36	0.0047	1.20	0.34	4.0	
	C 1546	0.058	0.11	0.011	0.003	17.78	53.28	18.87	2.93	0.36	0.0048	0.56	1.22	3.9	
	C 1547	0.058	0.11	0.011	0,002	17.55	53.54	17.35	2,96	0.37	0.0061	1.26	0.28	6.0	
K	C 1550	0.054	0.10	0.011	0.002	17.51	53,68	17.30	2.91	0.40	0.0063	0.68	1.05	5.5	
K	C 1551	0.059	0.14	0.011	0.003	17.23	53.68	18.30	2,99	0.63	0.0063	1.45	1.28	3.6	
	C 1552	0.054	0.11	0.009	0.002	18.02	51.99	20.71	2.86	0.36	0.0055	0.52	0.26	3	
	C 1553	0.046	0.12	0.009	0.001	17.80	53.34	17.10	2.98	0.44	0.0072	0.90	0.64	6.5	
	C 1554	0.061	0.12	0.007	0.001	17.86	52.38	19.95	2,95	0.36	0.0044	0.96	0.61	3.6	
K	C 1555	0.060	0.11	0.009	0.001	17.80	53.23	17.43	2.98	0.36	0.0060	1.45	0.64	5.3	
K	C 1556	0.054	0.10	0.011	0.002	17.56	53.59	17.17	2.99	0.36	0.0059	0.90	1.36	5.2	
K	C 1557	0.058	0.11	0.013	0.001	17.71	52.96	19.09	2.93	0.36	0.0054		0.06	5.3	

TABLE II

Tensile Properties of Alloy 718 Chemistry Variations with a 1750°F Solution Treatment

Heat Number	Ultimate Tensile Strength (psi)	0.2 % Yield Strength (psi)	Elongation (%)	Reduction of Area (%)		of Vari	able Elem
Condition: 175	50°F - 1 hour - a	ir cool + 1325°F	- 8 hours -	furnace cool	to 1150°	F - 8 ho	urs - air
Specification AMS	185,000	150,000	12	15			
5662 B						-	
KC 1539 KC 1540	197,300 197,700	154,300 137,600	21.3 13.9	49.7 20.9	0.91 0.94	0.68	5.41 5.41 5.34
KC 1541 KC 1542 KC 1543	195,600 208,100 208,100	163,900 162,400 166,300	12.5 12.7 8.7	27.9 27.3 11.5	0.37 0.97 1.37	0.74 0.75 1.12	5.47 6.47
KC 1544 KC 1545 KC 1546 KC 1547 KC 1550	198,900 186,400 191,000 220,800 206,900	164,700 119,900 148,700 198,300 185,300	10.7 18.6 15.5 8.0 13.8	24.1 30.6 23.9 22.2 28.4	0.52 1.20 0.56 1.26 0.68	0.31 0.34 1.22 0.28 1.05	6.03 4.03 3.94 6.00 5.93
KC 1551 KC 1552 KC 1553 KC 1554 KC 1555	191,200 171,300 212,200 167,300 206,700	122,100 122,600 177,200 92,700 171,400	26.7 23.5 10.7 33.7 9.3	46.5 55.5 23.7 52.8 20.0	1.45 0.52 0.90 0.96 1.45	1.28 0.26 0.64 0.61 0.64	3.85 3.94 6.53 3.64 5.31
кс 1556 кс 1557	199,400 207,400	163,000 185,400	8.1 9.4	13.0 22.4	0.90 0.94	1.36 0.06	5.25 5.17

TABLE III

Tensile Properties of Alloy 718 Chemistry Variations With a 1950°F Solution Treatment

Heat Number	Ultimate Tensile Strength (psi)	0.2 % Yield Strength (psi)	Elongation (%)	Reduction of Area (%)	Amount Ti	of Vari	able Eler
Condition: 19	50°F - 1 hour - a	ir cool + 1325°F	- 8 hours -	furnace cool	to 1150°	F - 8 hc	ours - ai:
Specification	180,000	150,000	12.0	15.0			
AMS 5664 a							
KC 1539	186,100	137,700	28,0	53.2	0.91	0.68	5.41
KC 1540	193,000	132,200	17.6	33,4	0.94	0.67	5.41
KC 1541	189,500	140,400	22.7	43.4	0.37	0.74	5.34
KC 1542	194,200	140,500	20.8	39.5	0.97	0.75	5.47
KC 1543	200,700	141,400	11.1	20.5	1.37	1.12	6.47
KC 1544	199,700	168,500	12.2	30.3	0.52	0.31	6.03
KC 1545	180,700	111,400	21.2	39.1	⇒ 1.20	0.34	4.03
KC 1546	190,700	140,500	21.7	41.7	0.56	1,22	3.94
KC 1547	206,700	179,300	13.9	44.2	1.26	0,28	6.00
KC 1550	192,800	144,300	24.0	42.4	0.68	1.05	5.93
KC 1551	178,200	114,600	32.9	54.7	1.45	1.28	3.85
KC 1552	153,200	102,900	29.3	60.1	0.52	0.26	3.94
KC 1553	203,800	170,300	11.1	29.1	0.90	0.64	6.53
KC 1554	170,500	98,700	34.6	56.0	0.96	0.61	3.64
KC 1555	199,700	146,800	16.7	27.3	1.45	0.64	5.31
кс 1556	198,400	145,400	10.9	29.5	0.90	1.36	5.25
KC 1557	197,800	159,900	17.6	49.8	0.94	0.06	5.17

TABLE IV

Tensile Properties of Alloy 718 Chemistry Variations With a 1750°F Solution Treatment and a 1500 Hour Exposure at 1200°F

Heat Numb	Ten	imate sile th (psi)	0.2 % Yield Strength (psi)	Elongation (%)	Reduction of Area (%		of Vari	able Elem
Condition:			air cool + 1325° rs - air cool	°F - 8 hours	- furnace co	ool to 1150°	PF - 8 h	ours - air
KC 1539 KC 1540 KC 1541 KC 1542	199 19 7 209	,600 ,600 ,300 ,900	173,600 152,200 156,600 173,300	18.8 10.1 12.7 11.7	46.0 18.9 23.0 23.1	0.91 0.94 0.37 0.97	0.68 0.67 0.74 0.75	5.41 5.41 5.34 5.47
KC 1543 KC 1544 KC 1545 KC 1546 KC 1547 KC 1550	192 188 196 223	,700 ,700 ,600 ,800 ,700	179,700 158,900 125,300 153,800 203,600 191,400	5.6 10.6 17.0 13.4 6.6 11.0	10.5 23.6 27.7 23.6 11.2 22.2	1.37 0.52 1.20 0.56 1.26 0.68	0.31 0.34 1.22 0.28 1.05	6.47 6.03 4.03 3.94 6.00 5.93
KC 1551 KC 1552 KC 1553 KC 1554 KC 1555	175 213 187	,200 ,800 ,600 ,100 ,500	137,500 127,500 182,000 128,500 198,600	23.3 21.2 10.8 19.9 7.9	40.3 56.9 22.0 37.7 13.4	1.45 0.52 0.90 0.96 1.45	1.28 0.26 0.64 0.61 0.64	3.85 3.94 6.53 3.64 5.31
KC 1556 KC 1557		,000 ,800	180,000 172,600	10.8 10.6	19.0 19.7	0.90 0.94	1.36 0.06	5.25 5.17

TABLE V

V-Notch Charpy Impact Strength of Alloy 718 Chemistry Variations

	Impact Strengt	h* (foot-pounds)	Amount	of Vari	able Eler	
Heat Number	1750°F Solution Treatment	1950°F Solution Treatment	Ti	Al	<u>Cb</u>	
Aging Condit	ion: 1325°F - 8 hours - fur	nace cool to 1150°F - 8 hours -	air cool			
KC 1539	34.3	56.8	0.91	0.68	5.41	
KC 1540	9.3	12.5	0.94	0.67	5.41	
KC 1541	11.1	24.1	0.37	0.74	5.34	
KC 1542	10.1	18.4	0.97	0.75	5.47	
KC 1543	8.1	9.1	1.37	1.12	6.47	
KC 1544	8.5	11.6	0.52	0.31	6.03	
KC 1545	13.1	24.8	1.20	0.34	4.03	
KC 1546	10.3	23.5	0.56	1.22	3.94	
KC 1547	8.5	21.8	1.26	0.28	6.00	
KC 1550	11.0	20.3	0.68	1.05	5.93	
KC 1551	36.0	67.1	1.45	1.28	3.85	
KC 1552	79.1	75.9	0.52	0.26	3.94	
KC 1553	7.8	11.1	0.90	0.64	6.53	
KC 1554	44.5	74.0	0.96	0.61	3.64	
KC 1555	6.5	17.9	1.45	0.64	5.31	
KC 1556	9.1	19.6	0.90	1.36	5.25	
KC 1557	10.6	36.9	0.94		5.17	

^{*} All data represent the average of two tests.

TABLE VI X-ray Diffraction Data of Residues Electrolytically Extracted from Alloy 718 Chemistry Variations in the Heat Treated Condition

Heat	10% Relative		n Metha		for•			Cl in W		r for	Amount	of Vor	riable Elem
Number	CbC							Ni Cb		3 101	Ti	Al	Cb
Conditio	n: 1750°	F - 1	hour -	air c	ool +	1325 ° F	- 8 h	ours -	furnace	cool	to 1150°	PF - 8 1	nours - ai
KC 1539	S		s						M		0.91	0,68	5.41
KC 1540	WV	S		S					M		0.94	0.67	5.41
KC 1541	M	S		M					M		0.37	0.74	5.34
KC 1542	M	S		W					M		0.97	0.75	5.47
KC 1543	W	S		M					M		1.37	1.12	6.47
KC 1544	М	s		VŠ			-		VW		0.52	0.31	6.03
KC 1545	M	M		VS					S		1.20	0.34	4.03
KC 1546	M	M		VS				***	W		0.56	1,22	3.94
KC 1547	M	M	VS				M	W			1.26	0.38	6.00
KC 1550	S	VS	M						M	• :	0.68	1.05	5.93
KC 1551	VS	W					 ,		s		1.45	1,28	3.85
KC 1552	vs	-					M	w	·		0.52	0.26	3.94
KC 1553	W	S	W		•				М		0.90	0.64	6.53
KC 1554	VS	W		VW					M		0.96	0.61	3.64
KC 1555	M	vs		W					М		1.45	0.64	5.31
KC 1556	s	VS	-	W				-	vs		0.90	1.36	5.25
KC 1557	VS	W	M	W			M	W			0.94	0.06	5.17

NOTE: S = strong M = medium W = weak* Trace amounts of CbC, Laves and M6C were also found in the residues produced by this solution.

TABLE VII

X-ray Diffraction Data of Residues Electrolytically Extracted from Alloy 718
Chemistry Variations in the Heat Treated and Exposed Condition

÷		•	10%	HCl i	n Meth	anol			10% HC	l in W	ater*				
	Heat	Re	lative	X-ray	Inten	sities	for:	Rela	tive X-r	ay Int	ensities	for:	Amount	of Vari	iable Ele
	Number		Cb C	Laves	Ni3C	o M6C			Ni3Cb	NixCb	γ'		Ti	Al	<u>Cb</u>
	**			***************************************							-				
			3 BC 0 0	 ÷				_							
	Conditio	n:	11200	F - I	hour -	air c	ool +	1325°	$\mathbf{F} - 8 \mathbf{h}$	ours -	furnace	cool	to 1150	°F - 8 h	ours - a
			+ 120	00°F -	1500 h	ours -	air	cool			and the second				
	KC 1539		M		w						M		0.91	0.68	5.41
	KC 1540		W	S		S					M		0.94	0.67	5.41
	KC 1541		M	M	M	M								- •	
	KC 1542		M	VS							S	:	0.37	0.74	5.34
						W					M		0.97	0.75	5.47
	KC 1543		W	VS		W					· VS		1.37	1.12	6.47
	KC 1544		W	М	W	M			٠.		М		0.52	0.31	6.03
	KC 1545		W	M	•	vvs									
	KC 1546		w M	M	37377.7						VS		1.20	0.34	4.03
					VVW	S			-		M		0.56	1.22	3.94
	KC 1547		W	W	M		•		· S	W	- 100-00		1.26	0.28	6,00
	KC 1550		М	S	W	*******			****		S	7	0.68	1.05	5.93
	*** ****						. •								
	KC 1551		S	W							vvs		1.45	1.28	3.85
	KC 1552		M		W				M	W			0.52	0.26	3.94
	KC 1553		W	VS	W						VW		0.90	0.64	6.53
	KC 1554	•	M	W		VW					vs		0.96	0.61	3.64
	KC 1555	•	M	VS		VW					VS		1.45	0.64	5.31
	KC 1556		М	VS		W					vvs		0.90	1,36	5.25
	KC 1557		W	W	. W	W			VS	W			0.94	0.06	5.17
												•		•	

NOTE: S = strong M = medium W = weak

Trace amounts of CbC, Laves and M6C were also found in the residues produced by this solution.

TABLE VIII

Multiple Regression Data for the Tensile Properties of Alloy 718 Chemistry Variation

Property	Multiple Regression Equation								
1950°F Solution Treatment									
Ultimate Tensile Strength	= -9755 + 107,400si + 46,500cb + 26,700Ti - 18,100sicb - 3,260cbTi - 2,090cb2								
0.2% Yield Strength	= $43,680 + 36,440$ CbAl + $10,000$ CbAl ² + $4,360$ Cb ² - $58,940$ Al - $7,150$ Cb ² Al - 525 Cb ² Al ²								
Elongation	= 33.4 + 95.9si + 21.8si ² cb + 1.6A1 - 98.9si ² - 24.5sicb - 2.1Ti								
Reduction of Area	= $132.5 + 2.0$ sicb + 0.8 cb ² - 37 si - 16.5 cb - 7.1 Ti - 6.8 A1								
1750°F Solution Treatment									
Ultimate Tensile Strength	= 5,375 + 54,850cb + 32,260A1 + 12,390A1 ² + 11,300Ti - 9,850cbA1 - 3,490cb ²								
0.2% Yield Strength	= $-685,200 + 407,700 \text{ cb} + 48,400 \text{ Al}^2 + 3,680 \text{ cb}^3 - 65,300 \text{ cb}^2 - 13,800 \text{ cbAl} - 4,360 \text{ Al}$								
Elongation	$= 235 + 45.2 \text{sicb} + 23.8 \text{si}^2 + 5.9 \text{cb}^2 - 171 \text{si} - 71.1 \text{cb} - 3.7 \text{sicb}^2$								
Reduction of Area	$= 160 + 37.7A1 + 33.4Si^2 + 2.9Cb^2 - 63.4Si - 37.2Cb - 33.2A1^2$								
1750°F Solution Treatment	with 1500 Hours Exposure at 1200°F								
Ultimate Tensile Strength	= 34,680 + 46,380cb + 39,440A1 + 16,530Ti - 5,730CbA1 - 3,090Cb ² - 2,500SiCb								
0.2% Yield Strength	= -155,700 + 99,370cb + 48,700Al + 15,150Ti - 7,940cbAl - 7,230cb ² -2,220Si	<u> </u>							
Elongation	= $82.6 + 94si^2 + 9.1sicb + 1.4sicb^2 - 111si - 15.1si^2cb - 11.7cb$								
Reduction of Area	= $193 + 5.1$ CbAl + 3.4 Cb ² - 45.1 Cb - 28.5 Al - 23.3 Si - 11.1 Ti								

TABLE IX

Multiple Regression Data for the Impact Properties of Alloy 718 Chemistry Variation

Multiple Regression Equation

Variation Acc

1950°F Solution Treatment

Impact Strength = $80 + 487\text{Si} + 112\text{Si}^2\text{Cb} + 4.4\text{Si}\text{Cb}^2 - 354\text{Si}^2 - 152\text{Si}\text{Cb} - 120\text{Si}^3$

1750°F Solution Treatment

Impact Strength = $260 + 475i^2 + 395iCb - 2955i - 37Cb - 11Ti - 8A1$

TABLE X

Multiple Regression Data for the Phase Relationships in the Alloy 718 Chemistry Variat

Multiple Regression Equation

As-Heat Treated - 1750°F Solution Treatment

= -146.7 + 138Ti + 51.2Cb + 22.2CbAl - 85Al - 24.4CbTi - 4.6Cb²

Laves = $507 + 466 \text{CbAl} + 43.3 \text{Si} + 23.1 \text{Cb}^2 - 1130 \text{Al} - 224 \text{Cb} - 44.9 \text{Cb}^2 \text{Al}$

 $M_{6}C = -192 + 1060si^{2} + 129cb + 17.2cb^{2}si - 555si - 175si^{2}cb - 16.9cb^{2}$

 $Ni_3cb = -143 - 167.6si^2 + 324si - 18.6siAl - 56.5Ti - 101.4sicb + 45.2cb$

 $cbc = -187 + 348si + 142cb + 22.6cb^2si - 194cbsi - 21.9si^2 - 17.7cb^2$

After 1500 Hours Exposure at 1200°F

 γ' = 16.1 + 101.7Ti + 97.7Ti² + 45.3CbAl - 106.8Al - 75.2TiAl - 41.3TiCb

Laves = $305 + 330 \text{CbAl} + 85 \text{Si} + 15.9 \text{Cb}^2 - 818 \text{Al} - 148 \text{Cb} - 30 \text{Cb}^2 \text{Al}$

 $M_{6}C = 197 + 251sicb + 1730si^{2} + 2.4cb^{2}si - 1540si - 280si^{2}cb - 6.2cb$

 $Ni_3Cb = -37 + 35.7Al^2 + 21Cb + 0.6CbTi - 26.4TiAl - 10.6CbAl - 7.5SiCb$

 $CbC = -9.8 + 116.7Al + 18.8Cb - 59SiAl - 22.4Al^2 - 6.1CbAl - 2.4Cb^2$

Electron Vacancy Number $(N_{V_{SS}})$

 $N_{V_{SS}} = 1.97 + 0.23Si + 0.03SiAl - 0.09Si^2 + 0.23Cb + 0.06Ti + 0.10Al$

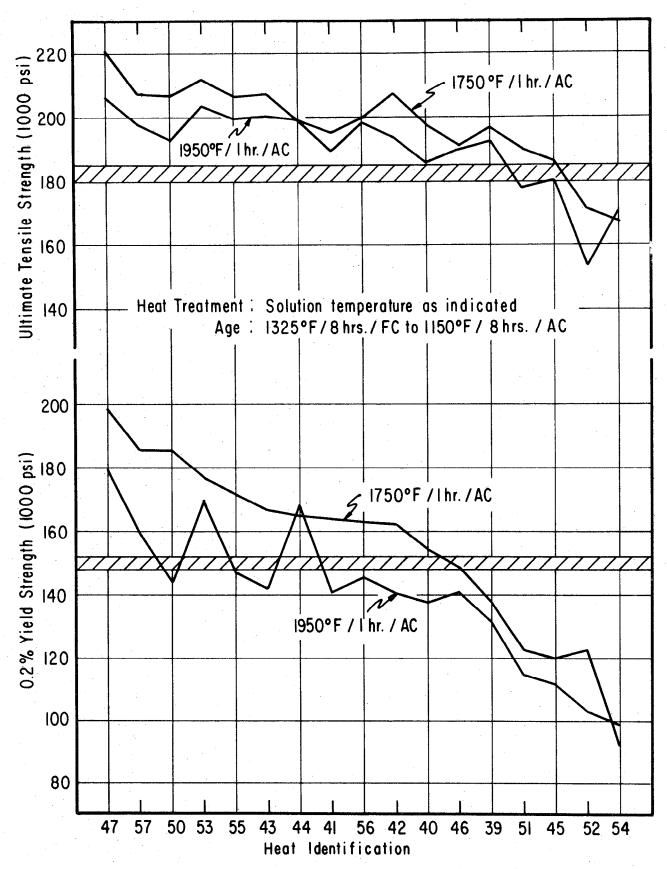


FIG. I: EFFECT OF SOLUTION TREATING TEMPERATURE ON ROOM TEMPERATURE ULTIMATE TENSILE AND 0.2% YIELD STRENGTH OF EXPERIMENTAL ALLOY 718

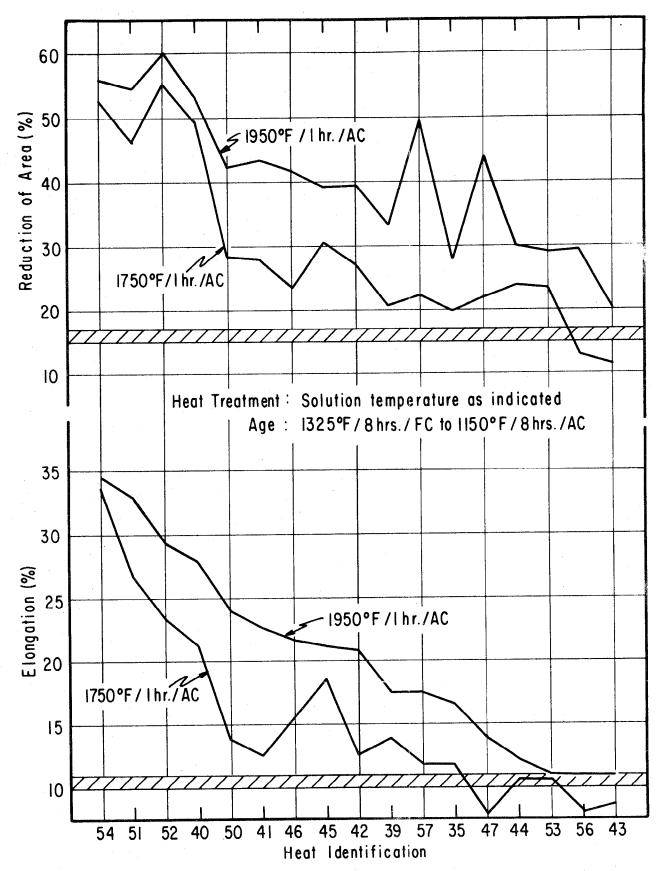


FIG. 2: EFFECT OF SOLUTION TREATING TEMPERATURE ON THE TENSILE DUCTILITY OF EXPERIMENTAL ALLOY 718 HEATS

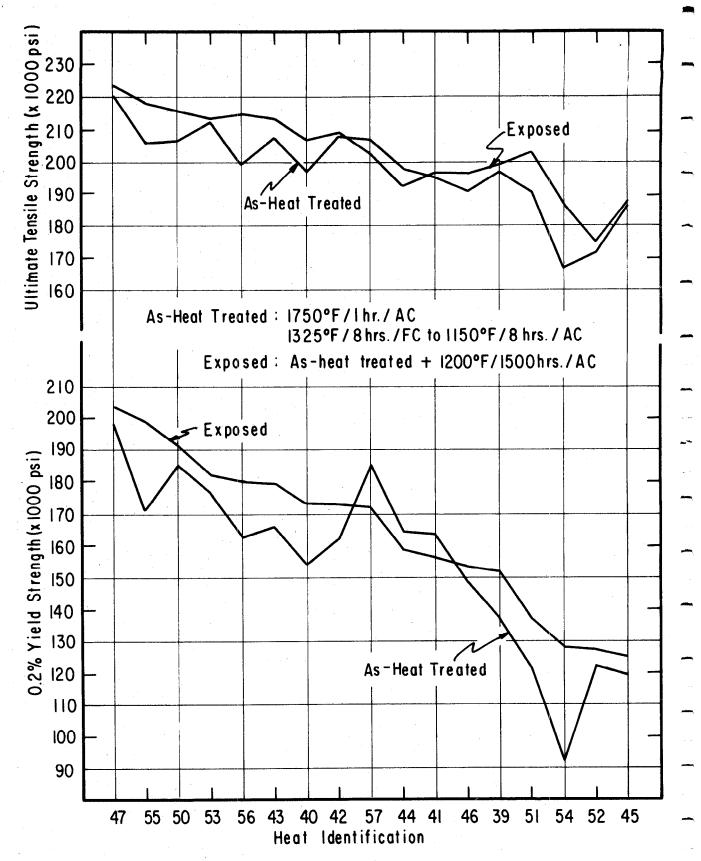


FIG. 3 : EFFECT OF LONG TIME EXPOSURE AT 1200°F ON THE --ULTIMATE TENSILE AND 0.2% YIELD STRENGTH OF EXPERIMENTAL ALLOY 718 HEATS

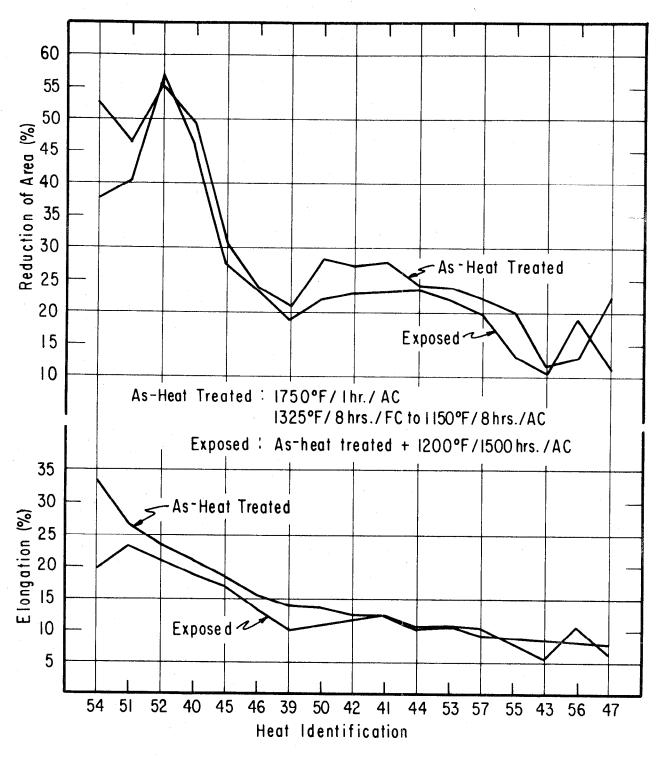


FIG. 4: EFFECT OF LONG TIME EXPOSURE AT 1200°F ON THE TENSILE DUCTILITIES OF EXPERIMENTAL ALLOY 718 HEATS

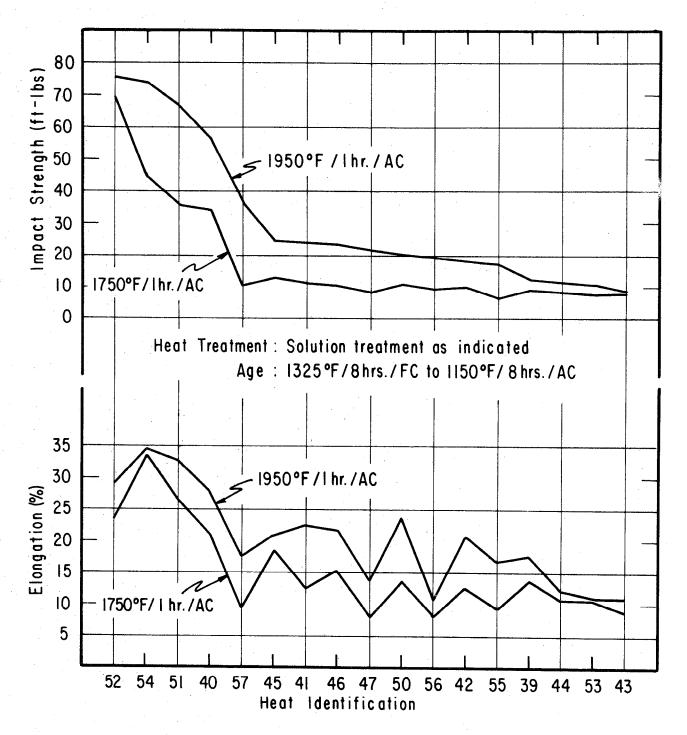


FIG. 5: EFFECT OF SOLUTION TEMPERATURE ON THE IMPACT STRENGTH OF EXPERIMENTAL ALLOY 718 HEATS AND ITS RELATIONSHIP TO TENSILE ELONGATION

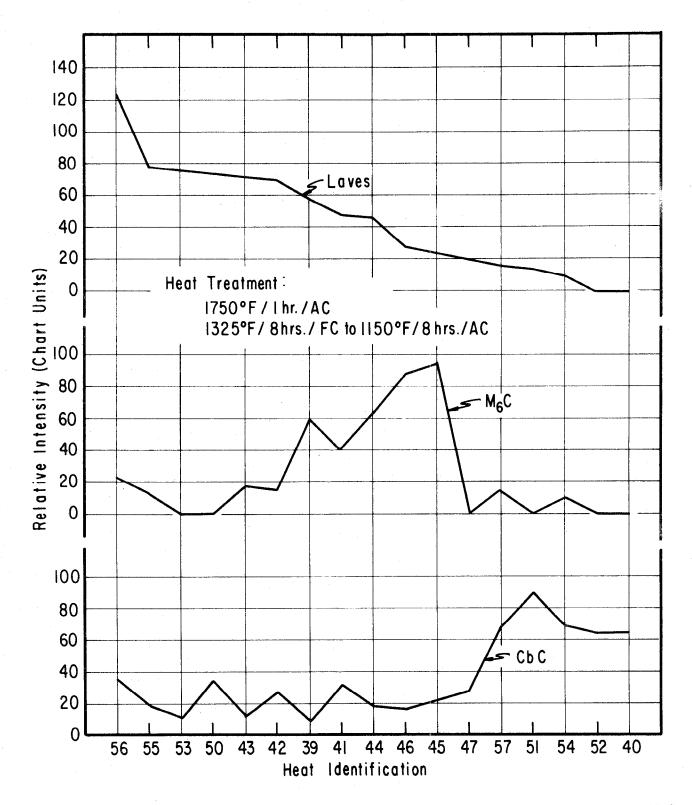


FIG. 6: RELATIVE AMOUNTS AND RELATIONSHIP OF LAVES, $M_6\text{C}$ AND CbC IN EXPERIMENTAL ALLOY718 HEATS

F 2 C

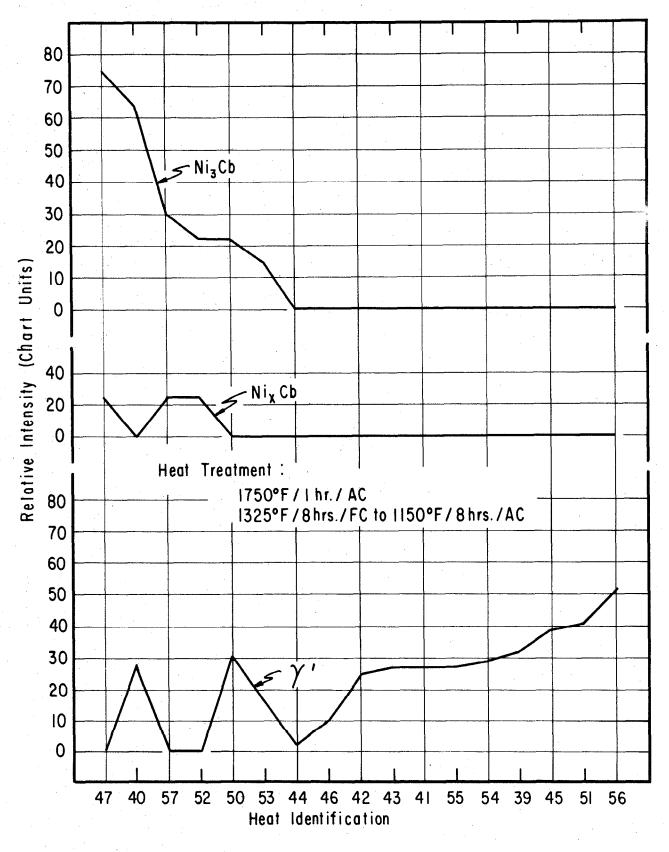


FIG. 7: RELATIVE AMOUNTS AND RELATIONSHIP OF Ni₃Cb, Ni_xCb AND γ' IN EXPERIMENTAL ALLOY 718 HEATS

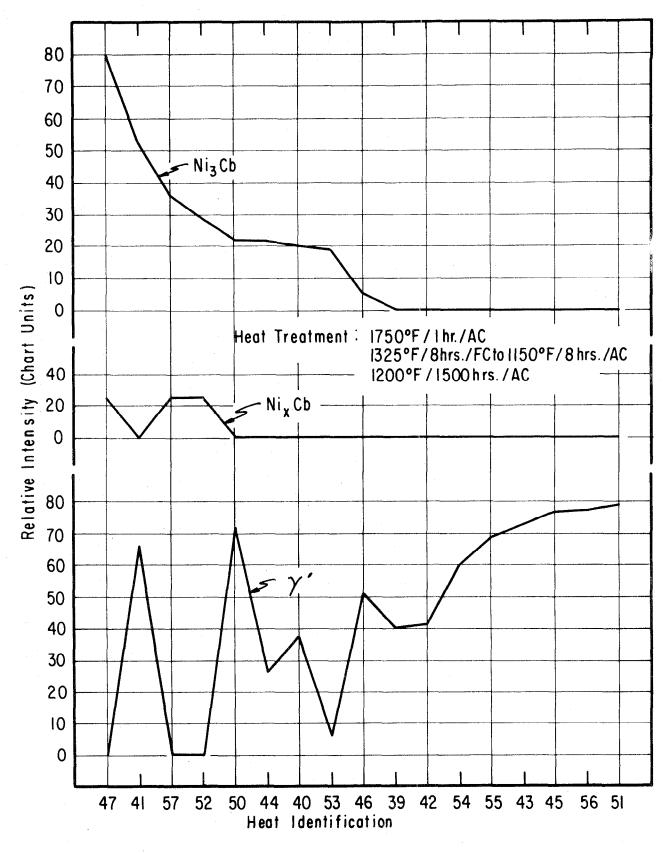
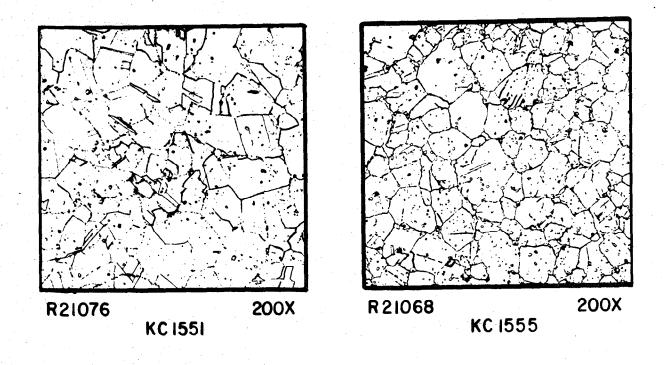


FIG. 8: RELATIVE AMOUNTS AND RELATIONSHIP OF Ni₃Cb, Ni_xCb AND γ' IN EXPERIMENTAL ALLOY 718 HEATS AFTER EXPOSURE AT 1200°F FOR 1500 HOURS



Heat Treatment:
1750°F/1 hr./AC
1325°F/8 hrs./FC to 1150°F/8 hrs./AC

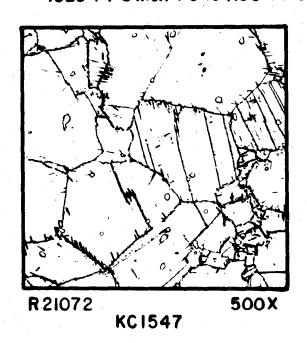
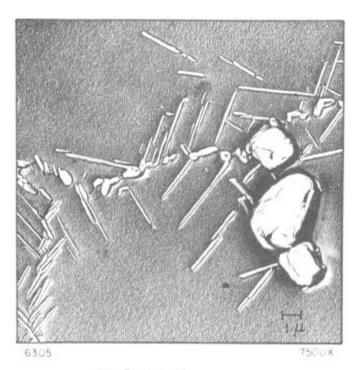
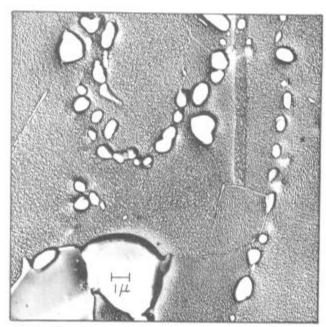


FIG. 9: LIGHT MICROGRAPHS ILLUSTRATING THE MICROSTRUCTURES OF SOME OF THE CHEMISTRY VARIATIONS OF ALLOY 718



1750°F / Thr. / AC 1325°F / 8hrs / FC to H50°F / 8 hrs. / AC

FIG. 10: ELECTRON PHOTOMICROGRAPH ILLUS-TRATING THE AS-HEAT TREATED MICRO-STRUCTURE OF HEAT KC1547

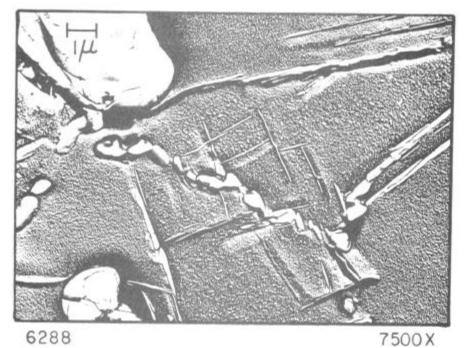


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1750°F / I hr. / AC 1325°F / B hrs. / FC to 1150°F / 8 hrs. / AC 1200°F / 1500 hrs. / AC

FIG. II: ELECTRON PHOTOMICROGRAPH ILLUS-TRATING THE MICROSTRUCTURE OF HEAT KC1544 AFTER EXPOSURE FOR 1500 HOURS AT 1200°F



1750°F / 1 hr. / AC 1325°F / 8 hrs. / FC to 1150°F / 8 hrs. / AC 1200°F / 1500 hrs. / AC

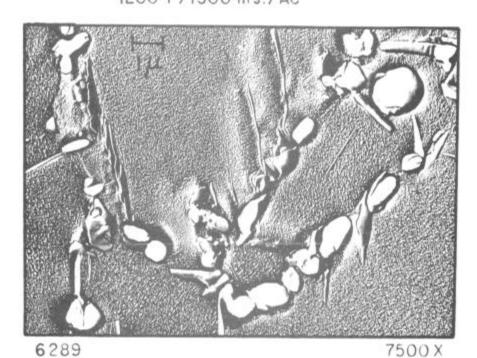
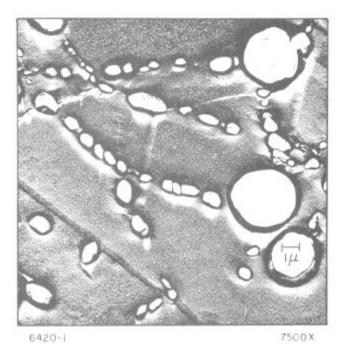


FIG. 12: ELECTRON PHOTOMICROGRAPHS ILLUSTRATING THE MICROSTRUCTURE EXHIBITED BY HEAT KC1547 AFTER 1500 HOURS EXPOSURE AT 1200°F

H-72138



1750°F / 1 hr / AG

 $1325^{\circ} \text{F/8} \, \text{hrs./FC}$ to $1150^{\circ} \text{F/8} \, \text{hrs./AC}$ $1200^{\circ} \text{F/1500} \, \text{hrs./AC}$

FIG. 13: ELECTRON PHOTOMICROGRAPH ILLUS-TRATING THE MICROSTRUCTURE EXHIB-ITED BY HEAT KC 1539 AFTER 1500 HOURS AT 1200°F

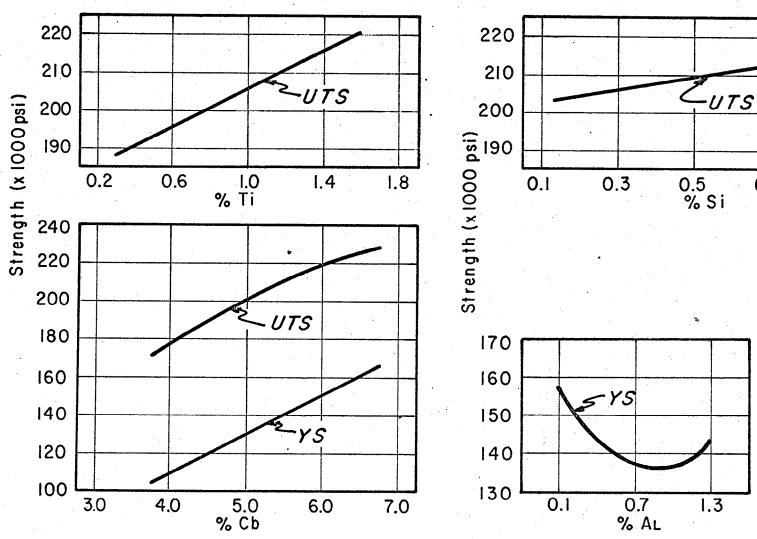
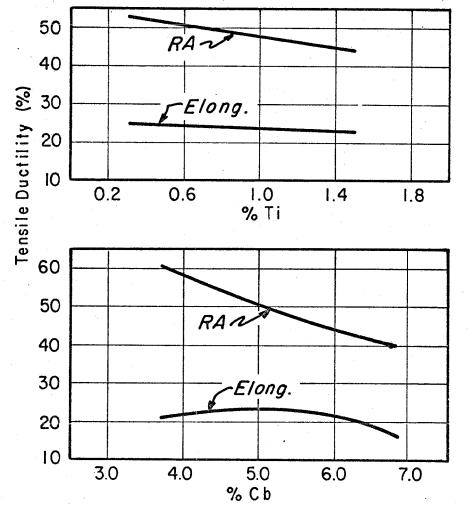


FIG. 14: REGRESSION CURVES SHOWING THE EFFECT Cb,Ti,Si AND AULTIMATE TENSILE AND 0.2% YIELD STRENGTH OF EXPERALLOY 718 HEATS SOLUTION TREATED AT 1950°F.



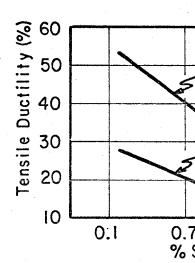


FIG. 15: REGRESSION CURVES SHOWING THE EFFECT OF Cb, SI AND T TENSILE DUCTILITIES OF EXPERIMENTAL ALLOY 718 HEA SOLUTION TREATED AT 1950°F



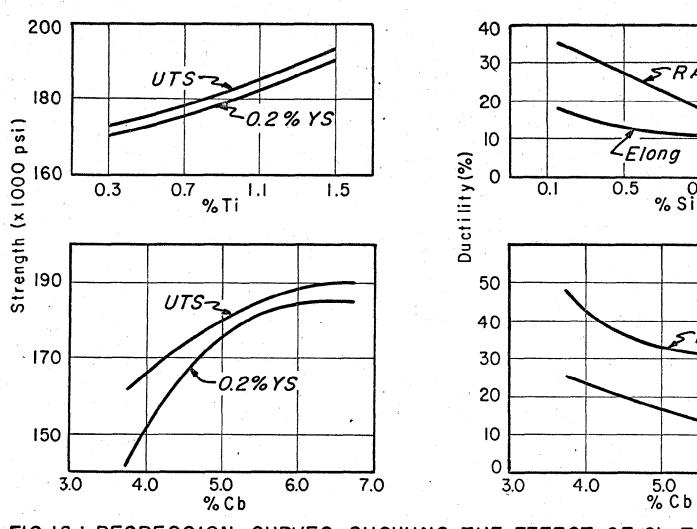
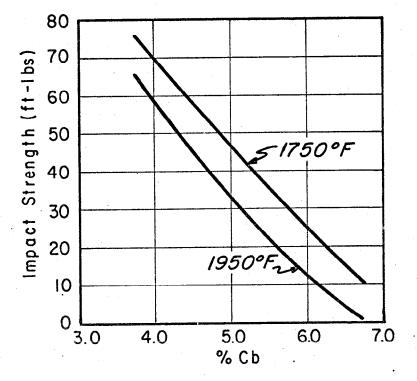


FIG. 16: REGRESSION CURVES SHOWING THE EFFECT OF Cb, T ON THE TENSILE PROPERTIES OF EXPERIMENTAL ALL HEATS SOLUTION TREATED AT 1750°F AND EXPOSED 1500 HOURS AT 1200°F



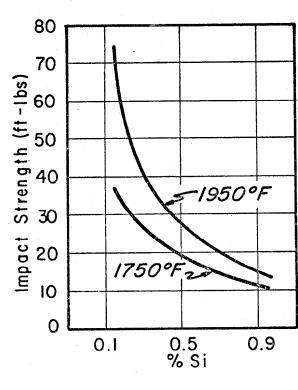
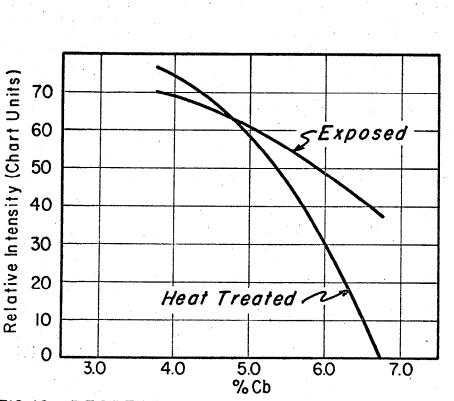


FIG. 17: REGRESSION CURVES SHOWING THE EFFECT OF Cb, SI AND SOLUTION TEMPERATURE ON THE IMPACT PROPERTIES OF EXPERIMENTAL ALLOY 718 HEATS



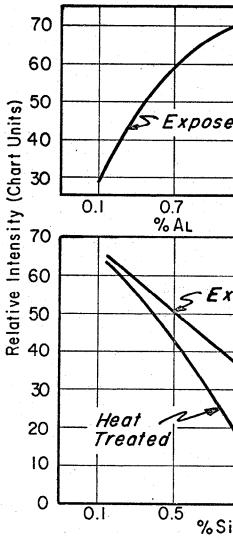
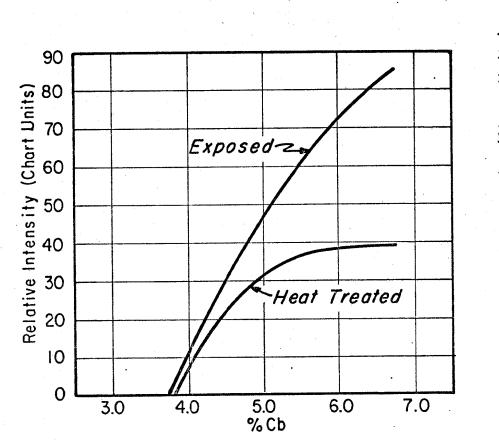


FIG.18: REGRESSION CURVES SHOWING THE EFFECT OF Cb,S; AND AL AMOUNT OF CbC PRESENT IN HEAT TREATED (1750°F SOLUTREATED) AND EXPOSED SAMPLES OF EXPERIMENTAL ALLO



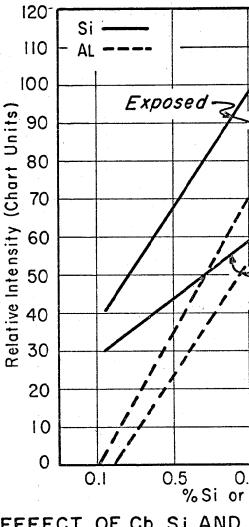


FIG. 19: REGRESSION CURVES SHOWING THE EFFECT OF Cb, Si AND AMOUNT OF LAVES PRESENT IN HEAT TREATED (1750°F SC TREATED) AND EXPOSED SAMPLES OF EXPERIMENTAL A HEATS.

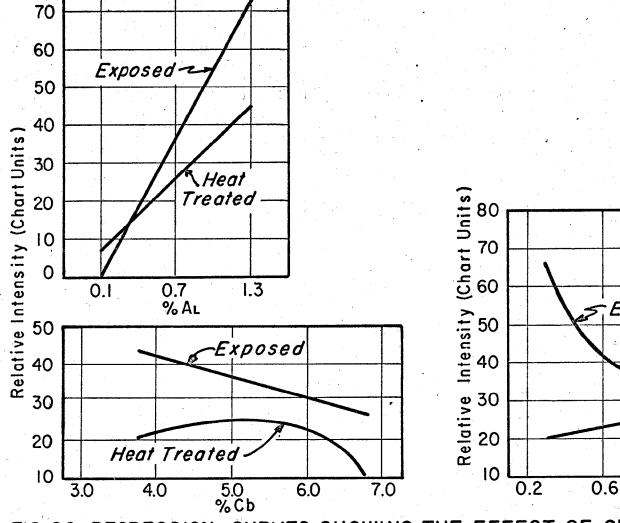


FIG. 20: REGRESSION CURVES SHOWING THE EFFECT OF Cb, Ti AND AL CAMOUNT OF Y' PRESENT IN HEAT TREATED (1750°F SOLUTION AND EXPOSED SAMPLES OF EXPERIMENTAL ALLOY 718 HEAT

Exposed

EHeat Trea

1.0 %Ti



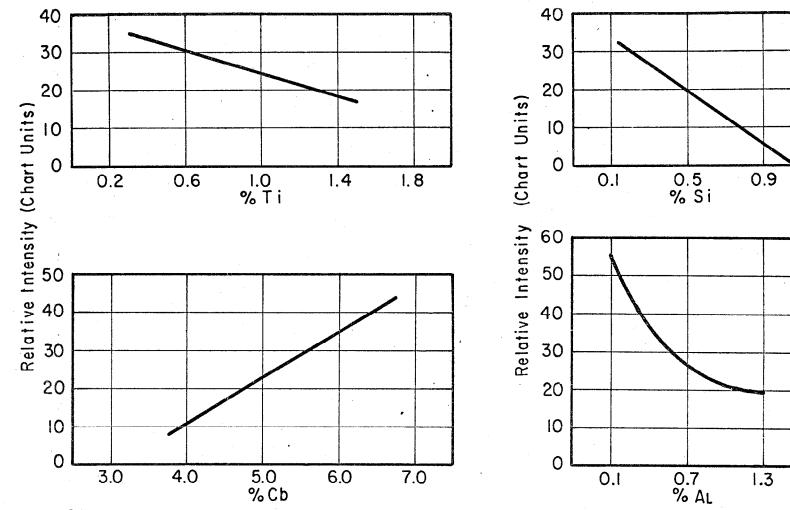


FIG. 21: REGRESSION CURVES SHOWING THE EFFECT OF Cb, AL, SI AND ON THE AMOUNT OF Ni₃Cb PRESENT ON SAMPLES OF EXPERIMENTAL ALLOY 718 HEATS EXPOSED AT 1200°F FOR 1500 HOUR