



ARAB REPUBLIC OF EGYPT
ATOMIC ENERGY AUTHORITY
DEPARTMENT OF METALLURGY

THE METALLURGY OF SUPERALLOYS
Part 1.

BY
M. P. ABDUL-AZIM AND E. H. HAMMAD

1990
INFORMATION AND DOCUMENTATION CENTER
ATOMIC ENERGY POST OFFICE
CAIRO, A.R.E.

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CONTENTS

	Page
ABSTRACT.....	i
THE METALLURGY OF SUPERALLOYS.....	1
1. Introduction.....	1
2. History of the Development of Superalloys.....	1
3. Structure and Phases of Superalloys.....	2
4. The Role of Alloying Elements in the Design of Superalloys.....	3
5. Superalloys Systems.....	5
6. Mechanical Properties of Superalloys.....	7
7. Summary.....	16
REFERENCES.....	18

ABSTRACT

This is part I of the report titled "The Metallurgy of Superalloys". In this part the structure, phases and systems of superalloys are reviewed. The role of alloying elements in the design of superalloys and the mechanical properties of superalloys are also reviewed.

Superalloys are important in high temperature technology, especially above 700°C. They are "super" mainly because their creep and stress rupture resistances are very high. Superalloys are based on an austenitic matrix including secondary phases, mainly γ' precipitates, inter and intragranular carbides mainly $M_{23}C_6$ and M_6C . They are classified into three systems, Ni-base, Fe-Ni base and Co-base alloys. Different alloying elements mainly Cr, Mo, Al, Ti are added to increase the strength either by solid solution hardening (Cr, Mo, Al), precipitation hardening (Al, Ti to produce γ') or by dispersion hardening (Cr, Mo to form $M_{23}C_6$ and M_6C carbides) and to increase the oxidation resistance (Cr, Al).

THE METALLURGY OF SUPERALLOYS

1. Introduction:

Superalloys are Ni-, Fe-Ni- and Co-base alloys generally used at temperatures above 550°C (823 K).

Superalloys have been developed for and utilized in a variety of high temperature applications such as gas turbines, aircraft turbine engines aircraft skins, spacecraft structures high temperature nuclear reactors, electrical power plants and chemical and petrochemical plants. Superalloys generally have the required properties for high temperature applications such as::

1. high creep resistance at application temperatures,
2. high resistance to thermal and mechanical fatigue,
3. high structural stability,
4. high resistance to hot corrosion in the working environment.

2. History of the Development of Superalloys:

In 1929 small amounts of Ti and Al were added to 80/20 Ni-Cr alloy, significant creep strengthening occurred. Superalloys were up. It was not until nearly 1940 that Bradely and Taylor envisioned the existence of the tiny coherent phase today referred as γ' which really caused the significant creep strengthening /1/. Despite inability to see such fine phase (γ'), the researches of the 1930s in England, USA Germany succeeded in creating strong alloys of Ni- and Fe-base containing Cr and γ' and carbide strengthened Co-base alloys.

The 1940s showed rapid changes and developments of superalloys driven by World War II. From these the further

improvement of superalloys and the invention of new alloy composition and new processes have been driven first by military use of jet engines and then by industry, which needed industrial gas turbines. Today although Al is most commonly used to achieve a high degree of precipitation hardening, Ti Ta and Nb are used extensively in combination with it to achieve optimum properties for specific uses or applications /1/.

3. Structure and Phases of Superalloys:

Superalloys consist of austenitic FCC matrix phase (γ) and a variety of secondary phases. The principal secondary phases are the carbides MC , $M_{23}C_6$, M_6C and M_7C_3 (rare) in all superalloy types and gamma prime γ' FCC ordered $Ni_3(Al,Ti)$ intermetallic compound in Ni- and Fe-Ni-base alloys/2,3/. In some limited systems where sufficient Nb or (Nb + Ta) is present, gamma double prime γ'' , body centered tetragonal is formed /2/.

In many systems unwanted or less desirable phases can be found owing to compositional, processing or exposure effects. Among these phases are sigma σ , delta δ orthorhombic Ni_3Nb , Laves, eta η HCP Ni_3Ti , mu μ and chi /2/.

Figure 1 is a panorama showing the 50-year development of Ni-base superalloy microstructure /1/. The figure is intended to characterize the microstructures that made the alloys increasingly strong and ductile over the years. However the lower part of the figure includes some of the problems causing phases in the microstructure that have been found to cause brittleness and/or lowered strength.

From the early 1930s until 1950s, the alloys created gradually were filled with more structure, this made them

increasingly stronger. However, this also led to the significant problems created by embrittling phases such as δ , μ and Laves. Generation of more and more γ' , both spherical and cubical, is apparent. The 1950s saw the creation of very complex grain boundaries, where carbides such as $M_{23}C_6$ lie engulfed in γ' , creating a dispersion-strengthened sheet bonding the grains together. By 1970, the effects of hafnium has been discovered and the γ' -engulfed carbide structure appeared unnecessary, since Hf controls the grain boundary to create strength and ductility in a more purely mechanical fashion.

Directional solidification processing creating aligned grain structures, aligned grain boundaries and even alignable filaments (such as TaC) shown as they appeared in the 1970s and 1980s. In 1980s the aligned homogeneous single-crystal structures appeared. Recently, through heat treatment transverse plates of γ' have been created in single crystals, which gives potential for still further strengthening /1/.

4. The Role of Alloying Elements in the Design of Superalloys

4.1. The base constituents of superalloys

For high temperature applications, the very first requirement is that the alloy should have the highest melting point. This means that the base constituent should be one of the refractory metals such as W, Mo, Nb, etc... However these metals have disadvantages. First, some of these elements tend to form volatile oxides at elevated temperatures /4/ which can result in run-away oxidation and metal loss. Second these elements have BCC structure which is inherently less tough and less ductile than FCC structure.

The next group of elements of higher melting point is iron, cobalt and nickel. Iron is the cheapest and most abundant of these three elements, but has BCC structure up to 910°C prior to transformation to FCC austenite (γ phase). An addition of 25 wt.% Ni to iron would stabilize the desirable FCC austenite even to temperatures below the ambient /5/. Hence, Fe-25 wt.% Ni base matrix is commonly used in many commercial superalloys.

Cobalt has a HCP structure up to 417°C prior to transformation to FCC structure. The FCC structure can be stabilized to lower temperatures by the addition of FCC stabilizing elements such as nickel /6/.

Since nickel has FCC structure at all temperatures it is used as the base constituent for superalloys /3/.

4.2. Alloying elements;

Superalloys contain a variety of elements in a large number of combination to produce the desired strength either by solid solution hardening, precipitation hardening or oxide dispersion hardening and the desired oxidation resistance.

Some alloying elements go into solid solution to provide one or more of the following: strength (Mo, Ta, W, Re, Ti, Nb, Va) /4, 7, 8/, oxidation resistance (Cr, Al) /8/, phase stability (Ni) /5, 6, 8/ or increased volume fractions of favourable secondary precipitates (Co) /8-11/. The solutes Ti, Al, Nb, Mo, Va, Ta, Ga, In, Ge and Si are added to form hardening precipitates such as γ' (Al, Ti) /4, 7, 8, 12/ and γ'' (Nb) /2/.

Minor elements (C,B) are added to form carbides and borides /4, 8, 13/, these elements plus others (Ce, Mg) are added for purposes of tramp element control /8,14/.

some elements (B, Zr, Hf) also are added to promote precipitation or carbide formation on the grain boundaries /4, 8, 17/.

The important carbide forming elements in superalloys are $Cr(M_{23}C_6, M_7C_3)$ Ti, Ta, Nb, Hf (MC) Mo and $W(M_6C)$ /8/.

All superalloys contain some Cr plus other elements to promote oxidation resistance. The presence of Cr is to promote the formation of Cr_2O_3 on the alloy surface. When sufficient Al is present, it promotes the formation of the more protective oxide Al_2O_3 /8/.

The function of alloying elements in Ni-, Fe-Ni- and Co-base superalloys is summarized in Tables 1-3 /8/.

5. Superalloys Systems::

The three types of superalloys, Fe-Ni, Ni- and Co-base are further subdivided into cast and wrought.

5.1. Fe-Ni base:

The most important class of Fe-Ni-base superalloys includes those alloys which are strengthened by intermetallic compound precipitation in a FCC matrix. The most common precipitates is γ' as in Incoloy 901, A-286 or V-57 but in some alloys γ'' is formed as in Inconel 718 /2,15/.

HNM and the CRMD series represent another class of Fe-Ni-base superalloys which are hardened by carbides, nitrides and carbonitrides/2,15/, elements such as W and Mo may be added to give solid solution hardening.

Other Fe-Ni-base superalloys consist of modified stainless steels primarily strengthened by solid solution hardening. Alloys in this class vary from 19-9DL (18 Cr-8 Ni with additional solid solution hardeners and higher C) to Incoloy 800H (32 Ni-21 Cr with small additions of Ti and Al which produces some γ' phase) /2/.

The Fe-Ni-base superalloys are used in the wrought condition although the CRMD series was developed primarily for casting applications.

The Fe-Ni-base superalloys have a strong tendency to form topologically close packed (TCP) phases such as sigma μ , Laves and chi /15/.

5.2. Ni-base.

Ni-base superalloys are the most complex, the most widely used for the hottest parts of all superalloys. Their use extends to the highest homologous temperature of any common alloy system; they are used over the temperature range 650-1100°C.

The most important class of Ni-base superalloys is that strengthened by intermetallic compound precipitation in a FCC matrix. The strengthening precipitate is typified by Waspaloy, Udimet 700, Inconel X-750 or Nimonic 80A /2,15/.

The second class of Ni-base superalloys are those which are strengthened by solid solution hardening but probably also by carbide precipitation produced through working and/or ageing schedule such as Inconel 617 or Hastelloy X /2/ .

Other Ni-base superalloys include oxide dispersion strengthened (ODS) alloys such as INMA-754 or INMA-600^{OE} /2/

Ni-base superalloys are used in both cast and wrought forms. Moreover wrought powder metallurgy alloys of the ODS class and cast alloys such as Mar-M247 have demonstrated property improvements owing to control of grain morphology by directional recrystallization or solidification /2/.

The Ni-base superalloys have a tendency toward precipitation of ordered geometrically close-packed (GCP) phases such as γ' and η /15/.

5.3. Co-base:

The Co-base superalloys are strengthened by a combination of carbides and solid solution hardening elements. The essential differences in these alloys is between cast and wrought structures. Cast alloys are typified by X-40 and wrought alloys by Haynes 25 /2/.

Co-base superalloys have not included an intermetallic compound possessing the same degree of utility as the precipitate in Ni-or Fe-Ni-base superalloys, although the alloy J-1570 has some of its strength from Ni_3Ti precipitation and certain high W content Co-base superalloys were hardened by precipitation of Co_3W /2/.

6. Mechanical Properties of Superalloys

6.1. Plastic flow resistance (yield strength)

The increase in yield strength by any strengthening mechanism depends on how effectively the strengthening microstructure can obstruct the motion of mobile dislocations. In dispersion strengthened or incoherent precipitation strengthened superalloys, the dislocations encounter the second phase particles and try to by-pass them by Orowan loop mechanism. Therefore the resulting yield strength σ depends on the critical Orowan stress required to achieve the by-pass process and is given by /4/:

$$\tau = \tau_0 + \alpha (G^{1/2} / l) \quad (1)$$

where τ_0 is the flow stress of the unstrengthened matrix which itself may have been increased by solid solution strengthening, G is the shear modulus of the alloy, b is the Burgers vector of the glide dislocations, l is the average interparticle spacing and α is a geometrical constant depending on the particle shape and distribution and the interparticle spacing. In addition to the Orwan contribution the inert dispersion strengthened alloys derive additional strength from the dislocation substructures introduced into these alloys during their thermomechanical processing.

In coherent precipitation strengthened superalloys such as γ' strengthened superalloys, the dislocations either cut the γ' precipitates or by-pass γ' particles when the interparticle spacing is large. As the interparticle spacing decreases, which can occur if either the volume fraction is increased or the particle size is decreased or both, the dislocations will start to cut through the coherent precipitates on the slip systems that are common to matrix and precipitate. In this case the superlattice dislocation pairs start to shear the ordered precipitates. When the coherent particles are very close together as in high volume fraction γ' superalloys, the resulting flow stress can be expressed by /4,16/.

$$\tau = (\gamma_{APB} / 2b) + [(\tau_0 + \tau_p) / 2] - T / br \quad (2)$$

where σ and σ_p are the flow stress of the matrix and the precipitate respectively, γ_{APB} is the antiphase boundary energy of ordered precipitate, T is the line tension of the dislocations and r is the equiaxed particle size. In equation (2) the major contribution (80%) to the flow stress comes from the first term which is the contribution of order strengthening. The coherency strain between the matrix and the precipitate can also contribute to strengthening, essentially by increasing the interaction distance between dislocations and particles /4/.

Orvan strengthening as in dispersion-strengthened superalloys such as TD-Nickel (Ni with 2% ThO₂) is very much effective in causing an increase in yield strength in comparison with coherent precipitation-hardened superalloys with high volume fraction of γ' . The major distinction between coherent precipitation and inert dispersion strengthened alloys is the temperature dependence of their strengthenings. Dispersion strengthening persists to near the solidus temperature because of the relatively non-reacting nature of the oxide dispersoids. On the other hand, coherent precipitation strengthening is temperature limited by thermally activated processes. For example all γ' strengthened Ni-base superalloys show a sharp fall-off in yield strength to solid solution levels at about 760-860°C which is hundreds of degrees below the γ' solvus temperature and at least one hundred degrees below the temperature where significant precipitate coarsening or dissolution can take place /4/. This fall-off may be due to slip caused disordering of the ordered precipitates resulting in a dynamic decrease in the antiphase boundary energy.

It is interesting to note that the most recent alloy development achievement, namely yttria dispersion strengthened superalloys, indeed combines the high strength qualities of precipitation strengthening at intermediate and ambient temperatures with very high temperature strength capability of inert dispersion strengthening /4/.

High strengthening can be caused with a uniform distribution of the particles where severe strain gradients can easily be produced. In the case of superalloys, the γ' coherent precipitates are uniformly distributed, while this is not the case in inert oxide dispersion strengthened alloys /4/.

Thus strengthening is increased if the volume fraction of the second phase particles is high, the particles are fine, the particle distribution is uniform and the particles are inert for high temperature applications.

6.2. Creep and stress rupture resistance:

Superalloys are "super" mainly because their creep and certain stress rupture resistances are very high. The Ni-base and Co-base superalloys are used in applications in the temperature range 600-1000°C. The temperature limit for TD-nickel, thoria and yttria dispersion strengthened superalloys is about 1150°C or slightly higher.

The overall steady state creep rate $\dot{\epsilon}_s$ of an alloy can be expressed for alloy design as a superposition of the creep rates due to dislocation or intragranular creep, grain boundary sliding creep and diffusional creep processes and is given by /4/:

$$\dot{\epsilon}_s = \dot{\epsilon}_{disl} + \dot{\epsilon}_{gbs} + \dot{\epsilon}_{diff} \quad (3)$$

Dislocation or intragranular creep is the result of the thermally activated motion of dislocations within the grains and it cannot be eliminated. The steady state creep rate due to dislocation creep can be expressed by /4/:

$$\dot{\epsilon}_{disl} = A \left[(\bar{\sigma} - \sigma_i) / E \right]^m \exp(-Q/RT) \quad (4)$$

where Q is the true activation energy for creep, m is the stress exponent of the steady state dislocation creep process, E is the young's modulus, A is a constant, $\bar{\sigma}$ is the applied stress and σ_i is the internal stress which must be overcome in the dislocation multiplication and glide processes.

Grain boundary sliding creep is a result of the accommodation of high shear strain gradients across grain boundaries in polycrystalline aggregates through sliding of adjacent grains along the grain boundaries. The creep rate due to grain boundary sliding can be expressed by /4/:

$$\dot{\epsilon}_{gbs} = Bb^3 \cdot D_B \cdot \sigma^2 / KTD \quad (5)$$

where d is the grain size of the alloy, D_B is bulk diffusion coefficient and B is a geometrical constant.

Diffusional creep is a result of the self-diffusion of atoms to the grain boundaries which are normal to the direction of the tensile stress. The steady state creep rate due to a diffusional flow can be expressed by /4/:

$$\dot{\epsilon}_{diff} = CD_3 \sigma b^3 \cdot 5 / K T d^n \quad (6)$$

where C is a geometric constant of the alloy and n is an exponent which is dependent on diffusion path, being 2 for bulk diffusion. This mode of creep is significant at very high temperatures (0.8 T_m). Although this mode may not produce much strain at lower temperatures, it can cause grain boundary chemistry changes leading to property degradation.

Equations (5) and (6) indicate that creep rates due to diffusional and grain boundary sliding processes can be reduced, if not eliminated by minimizing the grain boundary area per unit volume either by using single crystals or by very coarse-grained or alligand polycrystalline materials /4/.

Recent analyses of intragranular creep in terms of effective stress which is simply applied stress minus opposing average back stress are also tested for the applicability to grain boundary-related creep phenomena. The presence of grain boundary carbides in a Ni-base superalloy /17/ was found to increase this opposing back stress, indicating that grain boundary in this alloy system is either hindered or completely prevented by the presence of these precipitates.

The steady state creep rate in directionally recrystallized TD-nickel is 7 to 8 orders of magnitude smaller than the creep rate in nickel spite of their nearly equal activation energies which is that for vacancy self diffusion. This is because the creep stress exponent m is much larger in TD-nickel as compared to nickel. The steady state

creep rate in directionally solidified or monocrystalline Ni-base superalloys is very low. This is because in these alloys the stress exponent is large and the activation energy which is believed to be for interstitial self diffusion is nearly three times higher than that for vacancy self diffusion /4/.

The stress rupture life or ductility t_r can be expressed by /4/.

$$(\dot{\epsilon}_s)^\alpha \cdot t_r = \text{constant} \quad (7)$$

the exponent α is a function of the alloy system and microstructure and is found to be close 1 in many cases. According to equation (7) as the creep resistance increases i.e. $\dot{\epsilon}_s$ decreases, t_r increases. Unfortunately, in particles-hardened alloys, particles decrease $\dot{\epsilon}_s$ but decrease the stress rupture ductility. The stress rupture ductility for a Ni-20% Cr solid solution alloys is about 45%, for the precipitation strengthened Ni-base superalloy Udimet is about 8-12% and for the oxide dispersion strengthened superalloy TD-nickel is about 4-9% /4/.

Thus the creep rate can be greatly decreased and the stress rupture ductility increased if the material is monocrystalline, coarse grained or aligned polycrystalline, the volume fraction of particles is high, the particle distribution is uniform, the particles are fine and inert or are fine and coherent. For increasing stress rupture ductility and keeping the creep rate very low is still under investigation.

Pre-existing cracks can also propagate under creep loading conditions. The creep crack propagation is still under investigation. It was found that all conditions which reduce steady state creep rate and increase stress rupture ductility would improve the creep crack growth resistance /4/.

Figure 2 indicates the changes in superalloy creep rupture strength for 10000 h rupture life at 870°C and creep rupture ductility at 550-925°C with the issue of stronger superalloys /1/.

6.3. Ductility:

While the different modes of strengthening increase the strength of superalloys, they reduce their ductility even at high temperatures, in most cases.

It was found that /4/ not only particles reduce the ductility but also the interaction between plastic zones around particles can lead to greater reductions in ductility. When the volume fraction of particles is around 1% or less the particles behave as if they are isolated from each other. However, when the volume fraction of particles is about 10% interaction effects are significant and the interfacial stresses are enhanced due to the overlap of the plastic zones around particles. This results in interface separation at very low strains, which leads to the early fracture of the material as a whole. Even if the volume fraction is low, any local inhomogeneities of the particle distribution can result in localized decohesion and hence cause premature void nucleation and reduction in ductility. In addition incoherent interfaces, such those of oxide dispersoids are much weaker than the coherent interfaces, such those of γ'

precipitates and so lead to much reductions in ductility /4/

Phase instabilities resulting in the TCP and sigma phases and acicular carbides lead to reductions in ductility and toughness. Casting inhomogeneities, nonmetallic inclusions and shrinkage pores and voids can also lead to reductions in ductility and toughness. Trace or tramp elements such as tin, lead or arsenic even in ppm amounts can cause embrittlement in superalloys /4/.

The improvements of tensile ductility in particle strengthened alloys can be achieved when the volume fraction of particles is low, their distribution is uniform and the particles are fine, equiaxed and coherent with high particle-matrix interfacial strength /4/.

The ductility under more complex states of stress needs to be evaluated. Some studies /4/ show that the plane strain ductility of materials in general is less than their uniaxial ductility. Further, it appears that plane strain ductility decreases sharply with yield strength level than does the uniaxial ductility. There is very little understanding available at this time to explain why this is the case /4/.

7. SUMMARY

In the 1930s the strengthening of superalloys whether Fe-, Co- and Ni-base was just beginning. It soon became apparent that Fe- and Co-base alloys would be strengthened principally by solid solution and carbide strengthening, while Ni-base alloys would be strengthened by the coherent γ' precipitate. By 1950 the development of superalloys aimed to create more and more strength for higher and higher temperatures. Refractory metals (Mo, W, Nb and V) in increasing proportions were added to cause more and more solid solution and carbide strengthening. Al and Ti were added to create more γ' and Cr was decreased to increase the perceived allowable amount of γ' .

Interest concentrated in Ni-base superalloys, since (because of γ') they could be made stronger and also are natively more oxidation resistant. However, the increase in strength was generally accompanied by a concomitant decrease in ductility. By the 1960s widely used alloys such as In-100, Rene and B-1900 were testing the limits of ductility. Until 1970 the development of superalloys was limited by optimizing composition and heat treatment to have a high strength accompanied by, however, a considerable ductility.

By the middle 1970s, it became apparent that the critical failure mechanism at high temperatures was that of thermomechanical fatigue, a combination of ductility limitations in mechanical behaviour and effects from surface attack most succinctly expressed as a combination of low fracture toughness. Since it was not acceptable to recover some ductility by development dependent on alloy composition, superalloy metallurgists turned to developments in processing

(such as direction solidification and powder metallurgy).

Directional solidification process creates aligned grain structure, aligned grain boundaries and even alignable filaments (such as TaC) and single crystals. Powder metallurgy process creates oxide dispersion strengthened (ODS) superalloys. These developments in processing functioned to maintain or even improve strength properties while producing concomitant ductility. Now combined with γ' strengthening, wrought oxide dispersion strengthened superalloys generate usable creep-rupture strength to within 90% of melting temperatures.

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TABLE.1
EFFECTS OF SEVERAL ELEMENTS IN Ni-Base
SUPERALLOYS /8/

Element	Effect (a)
Chromium	Oxidation and hot corrosion resistance, solid solution strengthening
Molybdenum, tungsten	solid solution strengthening; form M_6C carbides.
Aluminium, titanium	Form $\gamma'Ni_3$ (Al, Ti) hardening precipitate; Ti forms MC carbides as well; Al enhances oxidation resistance.
Cobalt	Raises γ' solvus temperature
Boron, zirconium hafnium	Improve rupture life through increases in ductility, B also forms borides; Hf forms MC carbides and also promotes eutectic $\gamma-\gamma'$ formation in cast alloys
Carbon	Forms MC, M_7C_3 , $M_{23}C_6$ and M_6C carbides
Niobium	Forms $\gamma'' Ni_3Nb$ hardening precipitate, forms δ orthorhombic Ni_3Nb
Tantalum	Solid solution strengthening; forms MC carbides, enhances oxidation resistance.

a) Not all these effects necessarily occur in a given alloy.

Table . 2
EFFECTS OF SEVERAL ELEMENTS IN Fe-Ni-
Base SUPERALLOYS /8/.

Element	Effect (a)
Aluminium	Forms γ' Ni ₃ (Al,Ti), retards formation of hexagonal η Ni ₃ Ti.
Titanium	Forms γ' Ni ₃ (Al,Ti), forms MC carbides.
Niobium, tantalum	Form body centered tetragonal γ'' , form MC carbides, Nb forms δ orthorhombic Ni ₃ N b.
Carbon	Forms MC, M ₇ C ₃ , M ₂₃ C ₆ and M ₆ C carbides, stabilizes FCC matrix.
Phosphorus	Promotes general precipitation of carbides.
Nitrogen	Forms M (C,N) carbonitrides.
Chromium	Oxidation resistance, solid solution strengthening.
Molybdenum, tungsten	Solid solution strengthening; form M ₆ C carbides.
Nickel	Stabilizes FCC matrix, forms γ' and γ'' inhibits formation of deleterious phases.
Boron, zirconium	Improve rupture life through increase in ductility, retard formation of grain boundary η Ni ₃ Ti

a) Not all these effects necessarily occur in a given alloy.

Table . 3
EFFECTS OF SEVERAL ELEMENTS IN COBALT-
BASE SUPERALLOYS /8/.

Element	Effect (a)
Chromium	Oxidation and hot corrosion resistance; forms M_7C_3 and $M_{23}C_6$ carbides.
Molybdenum, tungsten	Solid solution strengthening; form M_6C carbides, form intermetallic compound CO_3M .
Tantalum, niobium	Solid solution strengthening; form M_6C carbides, form intermetallic compound CO_2M .
Aluminium	Oxidation resistance; forms intermetallic compound $CoAl$.
Titanium	Forms MC carbides, forms intermetallic compound Co_3Ti and with sufficient nickel Ni_3Ti
Nickel	Stabilizes FCC form of matrix, forms intermetallic Ni_3Ti , facilitates welding.
Boron, zirconium	Improve rupture life through increased ductility.
Carbon	Forms carbides MC , M_7C_3 , M_6C , $M_{23}C_6$

a) Not all these effects necessarily occur in a given alloy.

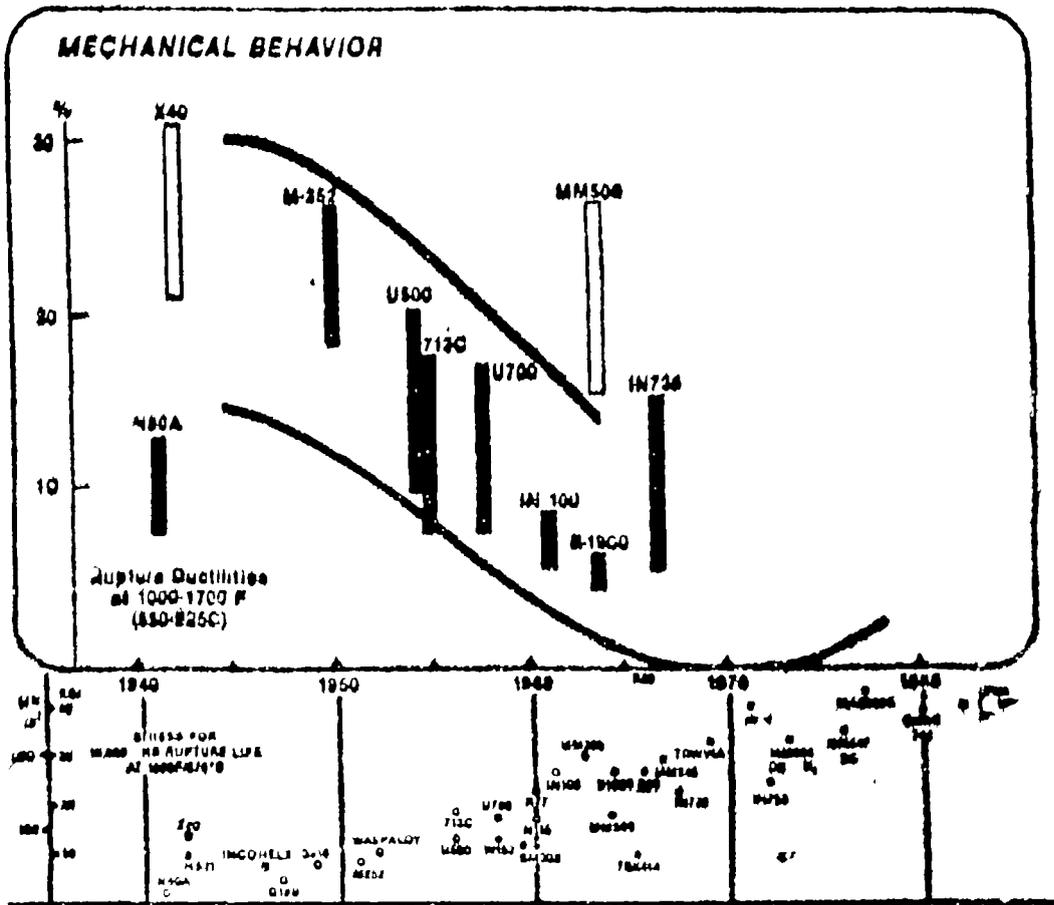


Fig. 2. An illustration of the change in superalloy creep-rupture ductility with the advent of stronger superalloys. /1/.

