



ARAB REPUBLIC OF EGYPT
ATOMIC ENERGY AUTHORITY
DEPARTMENT OF METALLURGY

THE METALLURGY OF SUPERALLOYS
Part 2.

M. B. ABD EL-AZIM AND E. H. HAMMAD

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CAIRO, A.R.E.

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BY

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ABSTRACT

This is Part II of the report titled "The Metallurgy of superalloys". It deals with the effect of heat treatment and operating conditions (thermal exposure and environment) on the mechanical properties of superalloys. The heat treatment is important in the development of superalloys through that it controls type, amount, size shape and distribution of the precipitate and the grain size of the matrix.

The thermal exposure leads to reduction in the amount of the primary carbides and to precipitation of secondary carbides. Also it leads to the agglomeration and coarsening of γ' or the transformation of γ' phase to γ phase. The environment may lead to the internal oxidation, carburization, decarburization or sulphidization of the superalloys which may result in the degradation of their mechanical properties.

This part gives also an example of applications of superalloys in the field of nuclear reactors especially high temperature-gas cooled reactors. Joined with this part a table which contains the major superalloys including its chemical analysis, creep rupture strength and some of its applications.

THE METALLURGY OF SUPERALLOYS

1. Introduction:

The development programme of superalloys includes the heat treatment of superalloys in order to improve its mechanical properties to be suitable for high temperature applications. The heat treatment controls amount and particle size and shape of the precipitate, carbide distribution and grain size.

During service superalloys are normally exposed to high temperature and may also be exposed to aggressive environments. The thermal exposure and the aggressive environment affect the mechanical properties of the superalloys and limit its service lifetime.

2. Effect of Prior Microstructure on Properties:

The principal microstructural variables of superalloys are:

- i) the precipitate amount and its morphology,
 - ii) grain size and shape and
 - iii) carbide distribution.
- Ni- and Fe-Ni-base superalloys properties are controlled by all three variables; the first variable is essentially absent in Co-base superalloys.

2.1 γ' precipitation:

Ni- and Fe-Ni-base superalloys may be strengthened by γ' precipitation in an FCC γ matrix. The γ' in Fe-Ni base superalloys and in Ni-base superalloys which contain

volume fraction (V_f) of γ' less than 0.25 generally is spheroidal, where in Ni-base superalloys which contain V_f of γ' higher than 0.35, γ' generally is cuboidal /1,2/. The volume fraction of γ' generally is about 0.2 or less in wrought Fe-Ni-base superalloys but may exceed 0.6 in Ni-base superalloys.

Strengthening by γ' is related to particle size, anti-phase boundary (APB) energy and stacking fault energy of γ' , strength of γ , strength of γ' , coherency strains, V_f of γ' , diffusivity in γ and γ' and γ - γ' modulus mismatch /2, 3-9/. However, the most direct correlations can be made with V_f of γ' and with γ' particle size.

Strength increases as V_f of γ' increases /1,8/, whereas ductility tends to decrease /1/. Although strength tends to increase as γ' size is reduced, in many instances the γ' size may be optimized in the range 0.2 - 0.5 μm /1/. Creep rupture strength is directly related to the V_f of fine γ' . The optimum creep rupture properties are produced by maximizing the V_f of fine γ' /1/.

2.2 Carbide precipitation-grain boundary hardening:

The nature of carbide hardening in superalloys is not nearly so well defined as it is for γ' . The role of globular or cellular grain boundary carbides has been determined in Ni- and Fe-Ni-base superalloys. The nature and extent of grain boundary hardening is not well identified in Co-base superalloys /1/.

Carbides exert a profound influence on properties by their precipitation on grain boundaries. In most superalloys, $M_{23}C_6$ forms at the grain boundaries after post casting or post solution treatment. It was found that /2/ a chain of discrete globular $M_{23}C_6$ carbides optimizes creep rupture life by preventing grain boundary sliding in creep rupture and providing sufficient ductility in the surrounding grain for stress relaxation to occur without premature failure. On the other hand, if carbide precipitate is a continuous grain boundary film, properties can be severely degraded. It was found that $M_{23}C_6$ films reduce the impact resistance of M252 /2/, and MC films lower the rupture life and ductility in forged Waspaloy /2/.

Another effect produced by grain boundary $M_{23}C_6$ carbide precipitation is the occasional formation of a zone depleted in γ' precipitate on either side of the boundary. These precipitate-free zones (PFZ) may have significant effects on the rupture life of Ni- and Fe-Ni-base superalloys. If such zones become wide or much weaker than the matrix, deformation would concentrate there, resulting in early failure. The superalloys of higher volume fraction of γ' do not show significant PFZ effects, probably because of their higher saturation with regard to γ' forming elements /2/.

Carbides at grain boundaries in Co-base superalloys act to inhibit grain boundary sliding and migration. In the highest C content cast Co-base superalloys, the carbide network may actually support a portion of load much as strengthening is achieved in a composite /1/.

2.3. Carbide precipitation-matrix or general hardening

The contribution of matrix carbide precipitation to the alloy strengthening is not well defined for any superalloy system.

The carbides formed within the grains affect the creep rupture strength of Co-base superalloys, and some Ni- and Fe-Ni-base superalloys /2/.

In Co-base cast superalloys script MC carbides are interspersed within the grains, causing a form of dispersion hardening which is not of a large magnitude owing to its relative coarseness /2/. The distribution of carbides in cast alloys can be modified by heat treatment, but strength levels at all the highest temperatures are substantially less than those of γ' -hardened alloys. Wrought Co-base superalloys have carbide modifications produced during fabrication sequence /2/. Solution treatments are not possible in most Co-base superalloys because melting occurs before all the carbides are solutioned /2/.

In Ni- and Fe-Ni base superalloys matrix carbides may be partially solutioned. Furthermore, MC carbides in these alloys tend to decompose to $M_{23}C_6$ at temperatures below about 816-871°C or possible convert to M_6C at temperatures of 982-1038°C if the alloy has a sufficient amount of Mo and W /2/. Matrix carbides generally contribute a very small increment of strengthening to Ni- and Fe-Ni-base superalloys.

2.4. γ'' precipitation

γ'' precipitation is restricted to Fe-Ni-base superalloys with Nb additions, as in Inconel 718 /2/.

The γ'' phase is not normally a stable phase. It can convert to γ' or to δ Ni_3Nb on long time exposure. The strength of γ'' is additive to that of γ' . Alloys hardened with γ'' achieve high tensile strength and very good creep-rupture properties at lower temperatures, but the conversion of γ'' to γ' or δ above about 677°C causes a sharp reduction in strength /2/.

2.5. Grain size

Grain size affects the strength of superalloys. A uniform grain size is often preferred, but is difficult to achieve in conventional forging operations or in large structural casting.

Coarse grain sizes improve the creep properties at high temperatures. Fatigue strength is affected by grain size, but the direction of improvement varies with the type of fatigue property evaluated, i.e., whether it is high cycle fatigue (HCF), low cycle fatigue (LCF) crack initiation or LCF crack propagation. Generally, for a fixed composition, HCF capabilities increase with finer grain size as does LCF crack initiation resistance. At elevated temperature LCF crack propagation generally increases with finer grain size /1/.

In a Ni-base superalloy In-792, it was found that increasing grain size markedly increases the crack growth resistance and so improves the toughness /10/.

The influence of grain size on ductility was studied in AISI 316 stainless steel /11/. It was found; that at high temperatures where fracture is essentially intergranular, the grain size has a pronounced effect on ductility, ductility decreases with increase in grain size. At low temperatures, fracture occurs by the transgranular ductile mode and high ductilities are obtained for all grain sizes. Ductility is not significantly influenced by grain size, although there is a general trend for ductility to increase with grain size to pass through a maximum and then to decrease for coarse grain sizes.

3. Effect of Thermal Exposure on Microstructure and Mechanical Properties:

The superalloys generally behave much like other alloys on thermal exposure during testing or in service but with some difference due to the nature of γ' precipitate. Most alloys with secondary phases undergo property degradation due to coalescence of secondary phases which reduces their effectiveness. The principal microstructure changes because of thermal exposure are : i) breakdown of primary carbides and formation of secondary carbides, ii) agglomeration of primary geometrically close-packed (GCP) strengthening phases such as γ' , and iii) formation of topologically close-packed (TCP) phases, such as sigma, Laves and mu.

3.1. Carbide morphology/type changes:

The thermal exposure leads to a reduction in the amount of primary MC carbides rich in Ti, Zr, Hf and refractory elements /2,12/. As exposure time increases, there is a tendency for further reduction in the amount of the primary

MC carbides and for further morphological changes together with a tendency to form increased amounts of the secondary carbides $M_{23}C_6$ or M_6C /2/. If carbide films form and/or acicular carbides of the M_6C type form, alloy ductility and strength can be reduced. Agglomeration of carbides, however, can lead to increased ductility accompanied by a reduction in strength. The precipitation of secondary $M_{23}C_6$ carbides intergranularly and intragranularly by ageing at 900°C for 1000 h led to a reduction in yield strength and an increase in ductility at low strain rates at 900°C as observed in Inconel 617 /13/. In the wrought Co-base solid solutioned alloy Haynes 25 (L-605), carbide precipitation at 815°C is responsible for alloy hardening in both early and late stages of exposure. In the late stages, $M_{23}C_6$ and M_6C participate in strengthening /14/. In Hastelloy X, extensive precipitation occurs at 705 to 790°C . As a result a dense intragranular secondary M_6C carbide forms and causes an increase in strength and a decrease in ductility. As primary M_6C carbides coalesce, there is a continual reduction in strength /14/.

Desirable high temperature mechanical properties are obtained from globular $M_{23}C_6$ carbides, while cellular $M_{23}C_6$ carbides reduce both rupture life and ductility as observed in Ni-base γ' strengthened Inconel X-750 and Ni-base solid-solution strengthened Inconel 600 by thermal exposure at 760°C /15/.

Superalloys normally used at lower temperatures of 425 - 725°C can be classified to two groups: i) stainless steels, ii) certain precipitation hardened Fe-base and Ni-base superalloys (for example : A-286, Incoloy 901, V-57, Waspaloy and Astroloy). These alloys may undergo microstructural changes

during testing or in service. If Ni-base alloys such as Waspaloy, Astroloy and Rene 95 are exposed for prolonged times at temperatures above 650°C, the strengthening phase may coarsen slightly. The same effect may occur above 600°C in Fe-Ni-base superalloys /14/.

Extensive changes in carbide phases can occur in austenitic stainless steels and may increase creep resistance. Carbide precipitation during exposure in solution-treated 316 stainless steel increases the creep resistance because strain ageing (dislocation-enhanced Cr_{23}C_6 precipitation) is very effective in retarding creep. Carbide precipitation during exposure could also influence fatigue life /14/.

There are two common means of reducing the severity of the carbide instability problem /12/. First, the addition of elements such as Hf, Nb and Ta can form more stable MC carbides than those of titanium, so that the driving force for $\text{MC} \rightarrow \text{M}_{23}\text{C}_6$ transformation and hence its reaction rate are reduced. However, this can only delay the problem and not eliminate it, since M_{23}C_6 is still the thermodynamically more stable phase. Second, a controlled heat treatments schedule could be used to precipitate the most stable M_{23}C_6 phase in discrete and chunky form at the grain boundaries.

Even the equilibrium carbide phase in superalloys can be prone to instabilities by adding complication of applied stress as a driving force /12/. This complication results from the increase in the chemical potential of vacancies at grain boundaries transverse to the stress axis to that of vacancies at boundaries nearly parallel to the stress axis. The only reasonable way to minimize this effect is to use either monocrystals or directionally solidified polycrystals instead of equiaxed polycrystals.

3.2 Phase instabilities of the γ' and γ'' :

The most common changes in γ' with exposure to elevated temperatures are coarsening and agglomeration under stress. Over heating can cause accelerated coarsening as well as solutioning of some fine γ' . Properties may be reduced in such cases, but when the overheating has been mild, precipitation of fine γ' occurs with a return to normal temperatures and some property recovery occurs /1/.

In Fe-Ni-base superalloys, extended exposure at 650-760°C can cause γ' to transform to η and γ'' to δ phases /12/. The driving force for the transformation of γ' to η is believed to be coherency strain energy associated with γ'/γ mismatch. η phase can form either as a cellular precipitate in the alternating η/γ lamellar arrangement emanating from the grain boundaries or as Widmanstätten platelets within the grains. Either morphology results in strength loss due to replacement of the fine γ' dispersion by the coarse η phase structure. The cellular η phase is also deleterious to stress rupture ductility. The driving force for the formation of η phase can be reduced by additions of those elements which are believed to reduce γ/γ' mismatch. Further boron additions apparently modify the grain boundary environs and suppress the cellular η formation in these alloys /12/.

In Fe-base superalloys with nickel ($\geq 45\%$) and high niobium ($\geq 5\%$), extended exposure at about 650°C can cause γ'' to transform to δ phase /12/. Addition of aluminum boron, replacement of niobium by tantalum and use of operating temperatures below 650°C are the common ways to avoid the $\gamma'' \rightarrow \eta$ transformation.

3.3. Formation of less-desirable phases

Superalloys may show a tendency to form less desirable TCP phases such as sigma, mu, Laves and Chi. These detrimental phases generally reduce property levels of the superalloys in which they form because of their inherent properties and/or the consumption of elements intended for γ and γ' /1/. These brittle phases can act as preferential paths for crack growth and, hence, cause reductions in ductility and toughness. In addition, they tie up a large number of refractory elements which could have otherwise acted as solid solution strengtheners in the superalloys.

Formation of TCP phases not only occur in Ni- and Co-base superalloys but also in Fe-Ni-base superalloys. Sigma and mu phases can be represented by the average formula $(CrMo)_x (NiCo)_y$ where x and y are usually equal and can vary from 1 to 7 /12/. σ tends to form from elements of nearly atomic size while μ from elements of somewhat greater atomic size differences. Laves phases, on the other hand, are A_2B compounds containing atoms bonded by contribution from size factors, rather than electronic effects /12/. These phases possess complex crystal structures but the most distinguishing feature of these phases is that their crystal unit cell contains close packed atom planes separated from each other by relatively larger interatomic distances /12/. Hence, the name topologically closed-packed phase as against the γ' or γ'' phases which are well packed in all directions and hence are referred to as geometrically close-packed phases /12/.

Hastealloy X may show a tendency to TCP phase formation above 750°C. Sigma (Ni-Cr-Mo) phase was identified in Hastealloy X exposed for more than 2500 h at 705 and 790°C.

μ phase was also identified at 705-955^oS /14/. In Incoloy 800H ζ phase especially lath-like shaped was formed after long time exposure at 650^oC and caused a reduction in strength and ductility /16-19/. In 316 stainless steel, Laves phase forms at 650 - 725^oC in 10 to 100 h. Sigma and Chi phases may form with extended exposure /14/.

Grain boundary embrittlement by μ and ζ phases was observed in 15 Cr/25 Ni and 15 Cr/40 Ni type Fe-base superalloys /20/. The embrittlement of Fe-base superalloys can be sensitive to small amounts of μ and ζ phases present at the grain boundaries, even when the weight fraction is between 0.0 and 1%. The degree of embrittlement was found to be directly proportional to the concentration coefficients of the μ and ζ phases at the grain boundaries. The embrittlement resulted in intergranular fracture during impact testing. This embrittlement effect was suppressed by reducing the grain size or by decreasing or removing the grain boundary embrittling μ and ζ phases by controlling the chemical compositions of the alloys /20/.

The formation of some TCP phases can be prevented by the application of compositional control guided by the electron vacancy number (N_V) concept. The formation of some TCP phases such as ζ , Laves and μ is found to be related to excess electron vacancies in the Fe, Ni and Co transition element base metals. By ascribing N_V values to the alloying elements of the γ matrix, a weighted N_V can be calculated for γ . By experiment, upper limits for N_V can be set for a given alloy composition to ensure the absence of TCP phases in reasonable exposure times /1/.

TCP phases can be avoided also by using powder metal lurgy because of the homogeneity that is intrinsic with these alloys made from foms (usually finer than 100 mesh) alloy powder /12/.

4. Environmental Effects:

Aggressive environments can cause severe problems during high temperature service. The environment can produce remarkable effects in deformation and fracture of superalloys at elevated temperatures. In some instances creep or fatigue lives can be shortened by several orders of magnitude.

Table (1) /21/ lists some of the principal environmental strengthening or weakening mechanisms. The list is not all inclusive, but attempt to identify the more significant mechanisms. There is often good evidence available that demonstrates the effectiveness of each of these mechanisms, but there is no ready way to predict which mechanism would dominate in a particular circumstance.

Several studies /21/ have demonstrated that the creep crack-growth rates of superalloys in aggressive environments can be 100 times greater than in vacuum or inert environments. Environmental effects appear particularly severe at intermediate temperature 600-700°C. Creep crack growth at these temperatures always appears to be intergranular, regardless of environment.

Fatigue properties are also environment sensitive. Fatigue-crack nucleation can be markedly accelerated by aggressive environments /21/..

The superalloys generally react with oxygen, and oxidation is the prime environmental effect on these alloys. At temperatures about 870°C and below general uniform oxidation is not a major problem. At higher temperatures oxidation is a great problem. The oxidation resistance of superalloys at temperatures below 980°C is a function of Cr content where Cr forms a protective oxide layer of Cr_2O_3 . At temperatures above 980°C the oxidation resistance is a function of Cr and Al contents. The latter element leads to the formation of protective Al_2O_3 oxide film. The higher the Cr content, the less Al may be required to form a highly protective Al_2O_3 layer /1/. Since the level of Al in many of alloys is insufficient to provide long term Al_2O_3 protection, protective coatings are applied. These coatings also prevent selective attack which occurs down grain boundaries and at surface carbides and inhibit internal oxidation or substructure interaction of O_2/N_2 with γ' envelopes, a process believed to occur in Ni-base superalloys /1/.

At lower temperatures about 870°C and below, accelerated oxidation may occur in superalloys through the operation of selective flaking agents. One of the accelerated oxidation processes is hot corrosion. The hot corrosion process is separated into two regimes: low temperature and high temperature. Hot corrosion can be resisted by the use of high Cr content (20 wt. %) in the alloy. This method is suitable for Co-base superalloys and many Fe-Ni-base superalloys and unsuitable for Ni-base superalloys especially those which should have high creep rupture strength because high Cr content is not compatible with high volume fraction of γ' /1/. Hot corrosion in high creep rupture strength Ni-base superalloys can be resisted by coating.

Exposure in air in the temperature range 900 to 1100°C produced a major loss in stress rupture life and ductility of a cast Ni-base superalloy IN738 /22/. The sensitivity to this environmental damage increased with decreasing test temperature in the range 1000 to 700°C. Oxygen was identified as the source of the damage for air exposure and indirect evidence supported grain boundary penetration of the gas to considerable depth. It was agreed that oxygen segregation can lead to grain boundary immobilization and unstable intergranular fracture at intermediate temperatures.

Carburization is one of the most common forms of degradation for superalloys, which are used in chemical and petrochemical processes, in heat treatment facilities and in high temperature gas cooled nuclear reactors. Since most of superalloys are based on iron and nickel in which carbon has high diffusivity and contain carbide forming elements such as Ti, Cr, Mo, Nb and W. Carbon can diffuse into these alloys and may strongly affect their mechanical, physical and chemical properties by enhanced carbide precipitation. Carburization contributes to embrittlement due to heterogeneous carbide precipitation along grain boundaries. A loss of ductility at low temperatures and even high temperatures makes the material susceptible to brittle fracture /13,23,26/.

Carburization during creep at 950°C in helium which contains traces of Co, CH₄ and H₂O accelerated the growth of surface cracking and led to loss in rupture life and ductility of Inconel 617 /27/. The effect of oxygen contents in helium on the creep properties of Inconel 617 was investigated at 1000°C /28/. The oxygen in helium causes a remarkable decrease of the creep rupture time. The effect of oxygen on

the creep rupture properties was attributable to the decarburization which tends to decrease the high temperature strength.

Hydrogen embrittlement at operating temperatures can occur in Ni- and Fe-Ni-base superalloys.

Sulphidizing environments are particularly aggressive. In one a superalloy a He-3% SO₂ mixture produced creep crack growth rate approaching 1 mm/min at 650°C /21/. With softer heat-resistant Ni-base superalloys sulphidizing atmospheres are less potent, but they are still more severe than air. Adding salt to sulfidizing environments, in the fashion of hot corrosion mixture, further promotes sulfidation damage. In pure nickel, sulphur forms an eutectic at 650°C and thus the embrittlement produced by sulphidizing environments might be postulated to be due to liquid films in grain boundaries or other local areas.

5. Applications of Superalloys:

5.1. Introduction:

Superalloys have been developed for and utilized in a variety of high temperature applications such as gas turbines, aircraft turbine engines, nuclear reactors, electrical power plants and chemical and petrochemical plants. The applications of superalloys in nuclear reactors especially high temperature gas-cooled reactor is reviewed below.

5.2. Nuclear Reactors

Superalloys generally are important in nuclear technology where some of these alloys are used in heat exchangers in different types of reactors. Fe-Ni-base superalloys are used in fast breeder reactor and Fe-Ni- and Ni-base superalloys are used in high temperature gas cooled reactors.

The high temperature gas cooled reactor (HTGR) is a graphite-moderated, helium-cooled nuclear system capable, in principle, of providing helium at temperature as high as 1000°C. HTGR is under intensive development in West Germany, USA and Japan because of its potential for electricity generation based on either a steam cycle or a direct cycle gas turbine and for supply of heat for chemical processes, such as the gasification of coal. In HTR for electricity generation using a direct cycle helium turbine and for supply of nuclear process heat the primary coolant temperatures are in the range 800 - 1000°C /29,30/.

In HTGR systems there are practical limitations on the purity of helium coolant that can be successfully maintained. The impurities, generally expected to be present

in helium are hydrogen (500 μ atom), methane (50 μ atom) and carbon monoxide (50 μ atom). Water may also be present, but usually in much lower concentrations, less than 0.5 μ atom /30/.

The high temperature metallic components for a HRI plant (HTR with direct cycle helium turbine) are : turbine and stator blades, hot segments between the blades in the turbine rotor and hot gas ducts and for PNP plant (prototype HTR plant for nuclear process heat) are: heat exchanger tubes, hot gas collector, baffle plates and reformer tubes /29/.

In HTGR for nuclear process heat the intermediate heat exchanger components operate up to 950°C with a planned service life of at least 100000 h. Although the primary stresses in normal service is below 5 N/mm², the components need to withstand temperature excursions and/or pressure transients that may occur under upset conditions. For thick-walled components, fatigue loads during start-up, shut-down power adjustments and upset conditions are airtically important. A specific requirement for the use of metallic materials in a HTR is resistance to corrosion in the primary coolant helium which contains traces of H₂O, CO and CH₄ which may lead to carburization or decarburization of the metallic materials /29/. Thus the metallic materials required for HTR heat exchanger components should have high creep resistance (high creep rupture strength, long term creep rupture life), high resistance to thermal and mechanical resistance and good oxidation and carburization resistance in the temperature range 800-1000°C.

The superalloys at present under consideration in HTGR are : Hastelloy X, Hastelloy S and Inconel 617 (Ni- based solid solution strengthened alloys) for hot gas ducts, Inconel 718 (Fe-Ni-based precipitation strengthened alloy) and Nimonic 80 A (Ni-based precipitation-strengthened alloy) for turbine blades and Incoloy 800H and Incoloy 802 (Fe-Ni-based solid solution strengthened alloys) for reformer tubes /29,31/.

6. SUMMARY

The prior microstructure affects the mechanical properties of superalloys. The principal microstructural variables of superalloys are the amount and particle size and shape of the precipitate, carbide distribution and grain size of the matrix. The optimum creep rupture properties of γ' precipitation strengthened superalloys are produced by high volume fraction of fine γ' . A chain of discrete globular carbides on the grain boundaries optimizes the creep rupture life by preventing grain boundary sliding. On the other hand, if carbide precipitate forms a continuous film on the grain boundary, the properties can be severely degraded. Coarse grain size improves the creep properties at high temperatures.

The thermal exposure affects the mechanical properties of superalloys through its effect on the microstructure. Thermal exposure leads to a reduction in the amount of the primary carbides and to precipitation of secondary carbides. If carbide films form and/or acicular carbides form alloy ductility and strength can be reduced. Desirable high temperature mechanical properties are obtained when globular carbides form. Thermal exposure leads to coarsening and agglomeration of γ' resulting in a decrease in the strength of the alloy. Thermal exposure may also lead to the transformation of fine γ' phase to coarse η phase resulting in loss in strength. One of the other effects of thermal exposure is the precipitation of less desirable phases such as ϵ and μ which result in the degradation of the properties of the alloy.

Aggressive environments can cause severe degradation of the properties of superalloys. The principal environmental weakening mechanisms are internal oxidation, carburization, loss of alloying elements to environment, formation of internal voids, pinning of grain boundaries and release of deleterious impurities.

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Table (1)

Environmental Strengthening or Weakening
Mechanisms /21/

Strengthening

Internal oxidation, carburization, etc
Load-carrying by corrosion products
Sealing of cracks by corrosion products
Increased crack closure, in fatigue
Crack-tip blunting or bifurcation

Weakening

Internal oxidation, carburization, etc
Loss of alloying elements to environment
Cracking or decohesion of corrosion products
Lowering of surface or grain-boundary energies
Formation of internal voids
Pinning of grain boundaries
Loss in load-carrying cross section
Prevention of crack-tip rewelding, in fatigue
Release of deleterious impurities.

Appendix I

Nominal Composition, Creep-Rupture Strength and Typical Applications of some Selected Superalloys (1-3)

Alloy	Nominal composition, %										Rupture strength at 1000 h, MPa		Characteristics and typical applications.					
	C	Ni	Fe	Co	Cr	Mn	Si	Mo	W	Nb	Ti	Al		Other	649 °C	816 °C	982 °C	1093 °C
NI-base Alloys :																		
Solid-solution hardening alloys :																		
Hastelloy B-2	0.02	Bal	2.0	1.0	1.0	1.0	0.1	28.0	—	—	—	—	—	—	—	—	—	Resists hot hydrochloric acid and chlorine gas.
Hastelloy N	0.06	Bal	6.0	0.2	7.0	0.8	1.0	16.5	0.5	—	0.50	0.35 Cu, 0.01 B	—	34	—	—	—	Resistant to hot molten fluoride salts.
Hastelloy S	0.02	Bal	1.0	2.0	15.5	0.5	0.4	14.5	1.0	—	0.2	0.02 La	262	68	—	—	—	Gas turbine parts, welding wire for high temperature components, high stability, low thermal expansion alloy.
Hastelloy X	0.10	Bal	18.5	1.5	22.0	1.0	1.0	9.0	0.5	—	—	—	234	69	14	4	—	Jet engine sheet parts, 3001 oxidation resistance.
Inconel 600	0.08	Bal	8.0	—	15.5	0.5	0.2	—	—	—	—	0.2 Cu	—	30	12	6	—	Heat treating and chemical processing equipments.
Inconel 617	0.10	Bal	—	12.5	22.0	0.5	0.5	9.0	—	0.3	1.2	0.2 Ni	274	37	25	10	—	Gas turbine aircraft engine parts.
Inconel 625	0.05	Bal	2.5	—	21.5	0.2	0.2	7.0	—	3.6	0.2	—	365	90	17	—	—	Aircraft engines and turbines, chemical processing.
Nimonic 75	0.12	Bal	2.5	—	19.5	—	—	—	—	0.4	0.15	0.25 Cu	—	—	—	—	—	—
Precipitation-hardening alloys :																		
Astrolloy	0.06	Bal	—	15.0	15.0	—	—	5.25	—	3.5	4.4	0.03 B	772	80	103	—	—	Forging for high temperature.
Gasalloy A	0.07	Bal	2.0	13.5	19.5	0.5	0.5	4.3	—	3.0	1.4	0.1 Cu, 0.05 B	507	176	—	—	—	Jet engine blades.
Inconel 718	0.04	Bal	18.5	—	18.5	0.2	0.2	3.0	—	5.1	0.9	0.15 C	593	—	—	—	—	Jet engine and rocket parts.
Inconel X 750	0.04	Bal	7.0	—	15.5	0.5	0.2	—	—	1.0	2.5	0.2 Cu	469	110	—	—	—	Gas turbine parts, bolts.

Cont.

Appendix I Cont.

Alloy	Nominal composition, %											Tensile strength at 1000 h, MPa			Characteristics and typical applications.				
												649	916	982		1079			
	C	Si	P	S	Mn	Cr	Co	Ni	Mo	Nb	Al	Other	MPa	MPa		MPa	MPa		
Mar-M 200	0.15	59	0.5	30.0	8.25	—	—	0.7	30.0	—	1.0	5.5	1.5 Ni, 3 Ti, 0.05 Mn, 0.015 B	—	—	—	—	—	
Fe-Ni-base Superalloys :																			
Solid solution hardening alloys :																			
Incoloy 800	0.05	32.5	Bal	—	21	—	—	—	—	—	0.38	0.38	—	—	—	—	—	—	Titanium creep forming discs, ethylene furnace tubes.
Incoloy 802	0.4	32.5	Bal	—	21	0.8	0.4	—	—	—	—	—	0.4 Cu	172	69	—	—	—	Furnace, heat exchanger parts.
Incoloy 800H	0.08	32.3	Bal	—	21	0.8	0.5	—	—	—	0.4	0.4	0.4 Cu	159	48	—	—	—	Low cost sheet, bar, forging alloy.
19-9 HL	0.5	9.0	Bal	—	19	1.1	0.6	1.25	1.2	0.4	0.3	—	—	255	59	—	—	—	
Precipitation hardening alloys :																			
A-286	0.05	26.0	Bal	—	15	1.4	0.4	1.25	—	—	2.15	0.2	0.05 V	317	75	—	—	—	Gas turbine parts, blades, bolts.
V-57	0.08	27.0	Bal	—	14.5	0.35	0.75	1.25	—	—	3.0	0.25	0.7 Ti, 0.01 S	483	—	—	—	—	Jet engine rotors.
Incoloy 901	0.05	22.7	Bal	—	13.5	0.4	0.4	6.20	—	—	2.5	0.2	—	441	76	—	—	—	Gas turbine rotors, parts.
Haynes 586	0.10	21.0	Bal	20	22.0	—	—	3.0	2.5	0.1	—	0.3	0.5 Ti, 0.02 Ti	—	—	—	—	—	
Co-base Superalloys :																			
Solid solution hardening alloys :																			
Haynes 23	0.10	10.0	9.0	Bal	20.0	1.5	0.5	—	15	—	—	—	—	372	107	28	—	—	Jet engine parts, sheets

Cont.

Appendix I Cont.

Alloy	Nominal composition, %													Rupture strength at 1000 h, MPa		Characteristics and typical applications	
	C	Mn	Fe	Co	Cr	Ni	Mo	Si	Al	Nb	Ti	Al	Other	648 MPa	982 MPa		
	0.10	22.0	3.0	Bal	22.0	1.25	0.3	—	18.0	—	—	0.08	1a	—	2a		9
Precipitation hardening alloys :																	
AR-213	0.15	—	—	Bal	19.0	—	—	—	4.7	—	3.5	6.5 Ta, 0.1 Ti	—	90	2a	—	Stents, tubing, resistance to hot corrosion.
MF-35N	—	35.0	—	35	20.0	—	10	—	—	—	—	—	—	—	—	—	Stress-corrosion resistant high strength castings.
Heat-resistant casting alloys :																	
X-40	0.50	10.0	1.5	Bal	25.0	0.50	0.5	—	7.5	—	—	—	—	138	138	55	Gas turbine parts, nozzle vanes.
J-1650	0.20	27.0	—	36	19.0	—	—	12	—	3.8	—	2 Ta, 0.02 B	—	—	—	—	High temperature steels.
MAR-M 918	0.05	20.0	0.5	Bal	20.0	0.20	—	—	—	—	7.5 Ta	—	—	138	22	—	High temperature steels.

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