

CHAPTER 6

Austenitic Stainless Steels

Summary

AUSTENITIC STAINLESS STEELS are the most common and familiar types of stainless steel. They are most easily recognized as non-magnetic. They are extremely formable and weldable, and they can be successfully used from cryogenic temperatures to the red-hot temperatures of furnaces and jet engines. They contain between about 16 and 25% chromium, and they can also contain nitrogen in solution, both of which contribute to their high corrosion resistance. Were it not for the cost of the nickel that helps stabilize their austenitic structure, these alloys would be used even more widely.

Introduction

Austenitic stainless steels have many advantages from a metallurgical point of view. They can be made soft enough (i.e., with a yield strength about 200 MPa) to be easily formed by the same tools that work with carbon steel, but they can also be made incredibly strong by cold work, up to yield strengths of over 2000 MPa (290 ksi). Their austenitic (fcc, face-centered cubic) structure is very tough and ductile down to absolute zero. They also do not lose their strength at elevated temperatures as rapidly as ferritic (bcc, body-centered cubic) iron base alloys. The least corrosion-resistant versions can withstand the normal corrosive attack of the everyday environment that people experience, while the most corrosion-resistant grades can even withstand boiling seawater.

If these alloys were to have any relative weaknesses, they would be:

1. Austenitic stainless steels are less resistant to cyclic oxidation than are ferritic grades

because their greater thermal expansion coefficient tends to cause the protective oxide coating to spall.

2. They can experience stress corrosion cracking (SCC) if used in an environment to which they have insufficient corrosion resistance.
3. The fatigue endurance limit is only about 30% of the tensile strength (vs. ~50 to 60% for ferritic stainless steels). This, combined with their high thermal expansion coefficients, makes them especially susceptible to thermal fatigue.

However, the risks of these limitations can be avoidable by taking proper precautions.

Alloy Families in Perspective

The fundamental criterion in the selection of a stainless steel is generally that it can survive with virtually no corrosion in the environment in which it is to be used. Good engineering practice sometimes requires that materials be selected for sufficient, but finite, service life. This is especially true for high-temperature service, for which creep and oxidation lead to limited life for all materials. The choice among the stainless steels that can be used in that environment is then based on the alloy from which the component can be produced at the lowest cost, including maintenance, over the intended service life. The ferritic stainless steels are less expensive for the same corrosion resistance but sometimes are found lacking because of:

- Lack of toughness, as is the case at subambient temperatures or in thicknesses greater than about 1.5 mm
- Lack of great ductility, specifically if more than about 30% elongation is needed

- Susceptibility to high-temperature embrittling phases when moderately alloyed

The less-expensive martensitic grades are used instead of austenitic when high strength and hardness are better achieved by heat treating rather than by cold work, and mechanical properties are more important than corrosion resistance. This is also the case for the more expensive PH grades, which can achieve corrosion resistance only matching the least corrosion resistant of the austenitic alloys.

Duplex grades match austenitic grades in corrosion resistance and have higher strength in the annealed condition but present the designer with challenges with regard to embrittling phases that can form with prolonged exposure to elevated temperatures and only moderate ductility like the ferritic alloys.

So, the austenitic grades are the most commonly used grades of stainless mainly because, in many instances, they provide very predictable levels of corrosion resistance with excellent mechanical properties. Using them wisely can save the design engineer significant costs in his or her product. They are a user-friendly metal alloy with life-cycle cost of fully manufactured products lower than many other materials.

The austenitic alloys can have compositions anywhere in the portion of the DeLong diagram labeled austenite shown in Fig. 1 (Ref 1). This diagram was designed to show which phases are present in alloys in the as-solidified condition, such as found in welds. Thus it also applies to castings and continuously cast products. As a practical matter of castability, the composition of most commercial alloys falls along the zone of several percent ferrite as cast. The salient feature of austenitic alloys is that as chromium and molybdenum are increased to increase specific properties, usually corrosion resistance, nickel or other austenite stabilizers must be added if the austenitic structure is to be preserved.

The traditional way of displaying the austenitic stainless steels is to present 302 as a base. Figure 2 shows one such diagram. Diagrams such as these treat alloys as an evolutionary family tree and subtly mislead. Many alloys were pushed toward obsolescence because of advances in processing. For instance, 321 was developed as an alloy in which the detrimental effects of carbon were negated by addition of titanium. The widespread adoption of the argon oxygen decarburization (AOD) in the 1970s made this alloy unnecessary, except for special circumstances, since carbon could be cheaply

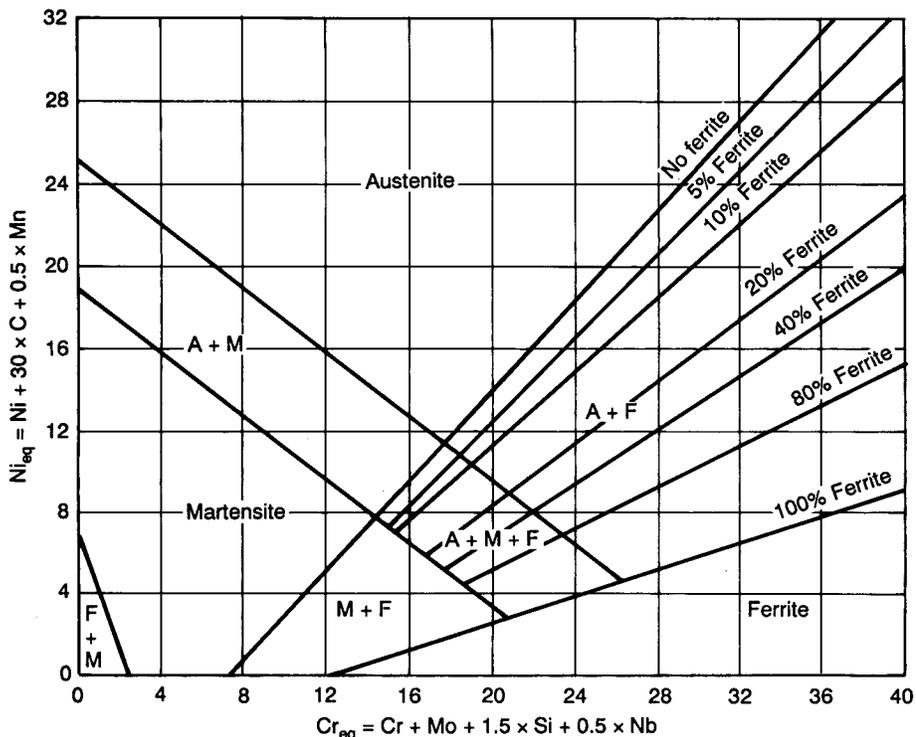


Fig. 1 Schaeffler-Delong stainless steels constitution diagram. Adapted from Ref 1, 2

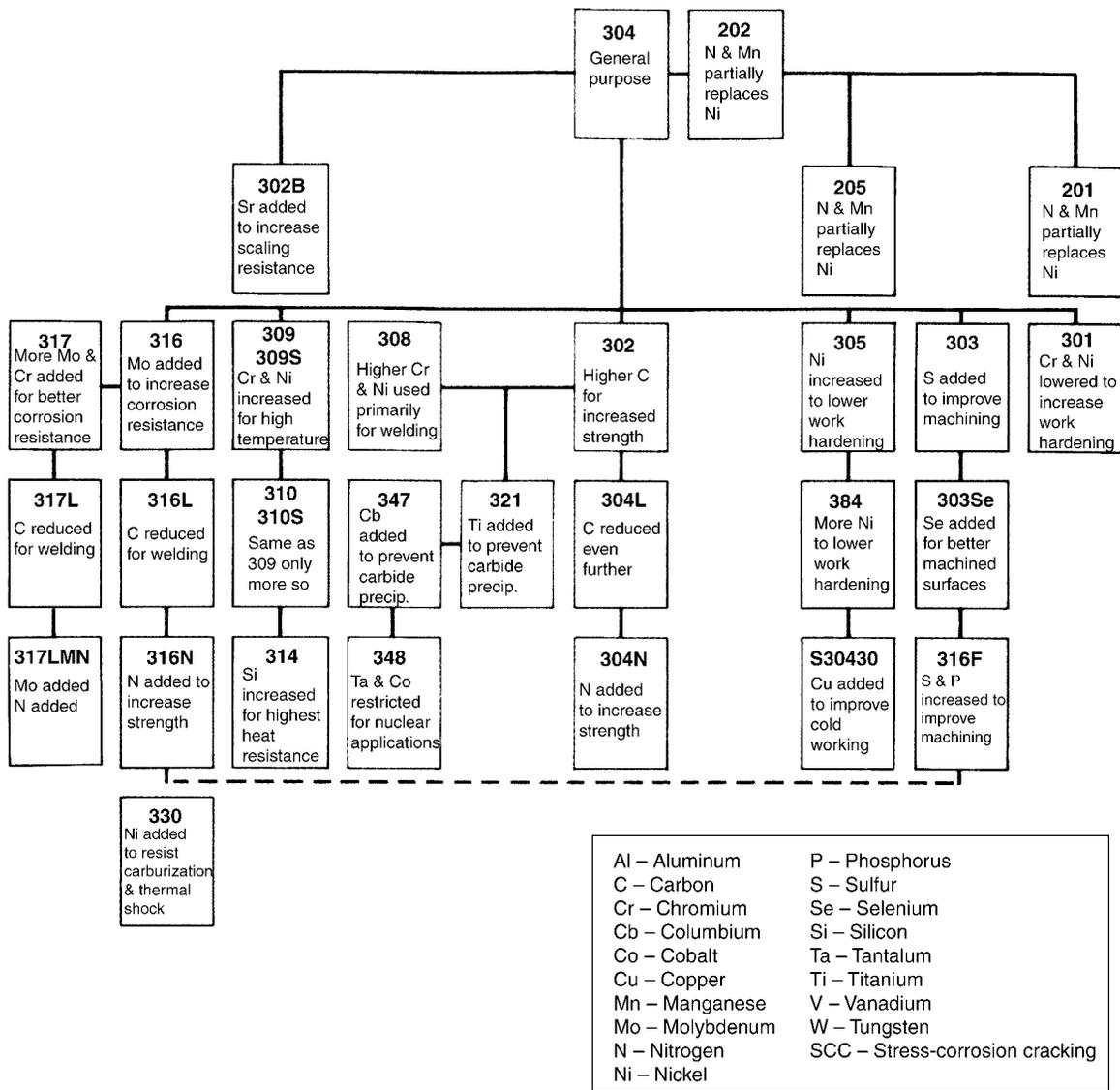


Fig. 2 The austenitic stainless family. Source: Ref 3

removed routinely. Likewise, 302 gave way to the lower-carbon 304, for which the even lower-carbon 304L is commonly substituted and dually certified to qualify as either grade. While low carbon prevents sensitization, stabilized grades may still be preferred for special applications such as type 321 in aerospace and type 347 in refinery service. Similar inertia keeps the higher-nickel 300 series as the de facto standard when the more cost-efficient high-manganese 200 series is the logical basic grade. The relevant types of austenitic alloys can nonetheless be rationalized with this diagram.

As chromium is added, oxidation resistance and corrosion resistance increase. Because

nickel equivalents (manganese, nitrogen, carbon, etc.) must also be added in matching amounts, austenite stability is also increased. If molybdenum, a chromium equivalent, is added, corrosion resistance but not oxidation resistance is enhanced. And, if nitrogen is the austenite stabilizer added to balance increases chromium or molybdenum, then corrosion resistance is also increased. With small exceptions, that is the rationale of austenitic grade design. Silicon is used as an alloy to promote oxidation resistance and resistance to corrosion by oxidizing acids. Copper is used to promote resistance to sulfuric acid. Rare earths make a more stably oxidation-resisting scale. Niobium increases

creep resistance. Sulfur and selenium increase machinability.

In this chapter, austenitic alloys are classified into three groups:

- Lean alloys, such as 201 and 301, are generally used when high strength or high formability is the main objective since the lower, yet tailorable, austenite stability of these alloys gives a great range of work-hardening rates and great ductility. Richer alloys, such as 305, with minimal work hardening are the high-alloy, lowest work-hardening rate grade for this purpose. The general-purpose alloy 304 is within this group.
- Chromium nickel alloys when the objective is high temperature oxidation resistance. This can be enhanced by silicon and rare earths. If the application requires high-temperature strength, carbon, nitrogen, niobium, and molybdenum can be added. 302B, 309, 310, 347, and various proprietary alloys are found in this group.
- Chromium, molybdenum, nickel, and nitrogen alloys when corrosion resistance is the main objective. Alloys such as silicon and copper are added for resistance to specific environments. This group includes 316L, 317L, 904L, and many proprietary grades.

Wrought alloys generally have cast counterparts that differ primarily in silicon content. Versions that require enhanced machinability have a high content of controlled inclusions, sulfides, or oxysulfides, which improve machinability at the expense of corrosion resistance. Carbon is kept below 0.03% and designated an L grade when prolonged heating due to multipass welding of heavy section (greater than about 2 mm) or when welds requiring a post-weld stress relief are anticipated.

Lean Alloys

Lean austenitic alloys constitute the largest portion of all stainless steel produced. These are principally 201, 301, and 304. Alloys with less than 20% chromium and 14% nickel fall into this unofficial category. Since they are stainless, it is generally taken for granted that these alloys will not corrode, and these alloys have sufficient corrosion resistance to be used in any indoor or outdoor environment, excluding coastal. These grades are easily weldable and formable and can be given many attractive and useful surface finishes, so they are very much general-purpose alloys. Table 1 lists some typical compositions of the most commonly used lean austenitic alloys. These typical compositions vary with end use, raw material cost factors, and the preference of a given manufacturer. The compositions of standard alloys are often fine-tuned to the intended end use. In this table, the word *drawing* indicates higher nickel for lower work hardening, while *tubing* indicates alloys with higher sulfur to facilitate gas tungsten arc welding (GTAW) penetration. *Tensile* indicates lower alloy levels to increase the work-hardening rate for material that is intended to be used in the cold-worked, high-strength condition. 316L is included in its most common tubing end use chemistry even though it is a corrosion-resisting alloy because it is so pervasively used as a service center sheet item.

The main difference among the lean austenitic alloys lies in their work-hardening rate: the leaner the alloy, the lower the austenite stability. As unstable alloys are deformed, they transform from austenite to the much harder martensite. This increases the work-hardening rate and enhances ductility since it delays the onset of necking since greater localized

Table 1 Typical compositions of the most commonly used lean austenitic alloys

Alloy	Designation	C	N	Cr	Ni	Mo	Mn	Si	Other	Other	Other
201	S20100	0.08	0.07	16.3	4.5	0.2	7.1	0.45	0.001 S	0.03 P	0.2 Cu
201 drawing	S220100	0.08	0.07	16.9	5.4	0.02	7.1	0.5	0.001 S	0.30 P	0.6 Cu
201LN	S20153	0.02	0.13	16.3	4.5	0.2	7.1	0.45	0.001 S	0.03 P	0.5 Cu
301 tensile	S30100	0.08	0.4	16.6	6.8	0.2	1.0	0.45	0.001 S	0.03 P	0.3 Cu
301 drawing	S30100	0.08	0.04	17.4	7.4	0.02	1.7	0.45	0.007 S	0.03 P	0.6 Cu
303	S30300
304	S30400	0.05	0.05	18.3	8.1	0.3	1.8	0.45	0.001 S	0.03 P	0.3 Cu
304 drawing	S30400	0.05	0.04	18.4	8.6	0.3	1.8	0.45	0.001 S	0.03 P	0.3 Cu
304 extra drawing	S30400	0.06	0.04	18.3	9.1	0.3	1.8	0.45	0.001 S	0.030 P	0.4 Cu
304L tubing	S30403	0.02	0.09	18.3	8.1	0.3	1.8	0.45	0.013 S	0.030 P	0.4 Cu
305	S30500	0.05	0.02	18.8	12.1	0.2	0.8	0.60	0.001 S	0.02 P	0.2 Cu
321	S32100	0.05	0.01	17.7	9.1	0.03	1.0	0.45	0.001 S	0.03 P	0.4 Ti
316L	S31603	0.02	0.0	16.4	10.5	2.1	1.8	0.50	0.010 S	0.03 P	0.4 Cu

deformation is more than offset by greater localized strain hardening.

These grades are best viewed as a continuum with a lower boundary at 16%Cr-6%Ni and an upper boundary at 19%Cr-12%Ni. This represents the range from minimum to maximum austenite stability. Since that is the main distinction within this grade family, let us examine its basis.

Martensite and Austenite. Stability. The formation of martensite at room temperature may be thermodynamically possible, but the driving force for its formation may be insufficient for it to form spontaneously. However, since martensite forms from unstable austenite by a diffusionless shear mechanism, it can occur if that shear is provided mechanically by external forces. This happens during deformation, and the degree to which it occurs varies with composition according to (Ref 4):

$$M_{d30} (\text{°C}) = 551 - 462(\%C + \%N) - 9.2(\%Si) - 8.1(\%Mn) - 13.7(\%Cr) - 29(\%Ni + Cu) - 18.5(\%Mo) - 68(\%Nb) - 1.42(GS - 8) \quad (\text{Eq 1})$$

This is the temperature at which 50% of the austenite transforms to martensite with 30% true strain (Ref 5). It should be noted that even elements that are chromium equivalents in promoting ferrite are austenite stabilizers in that they impede martensite formation. This temperature is the common index of austenite stability. This regression analysis was generated for homogeneous alloys. If alloys are inhomogeneous,

such as occurs when they are sensitized or when solute segregation occurs, as from welding, then the equation applies on a microscopic scale. Sensitized zones (i.e., the regions near grain boundaries where chromium carbides have precipitated) will have a much higher tendency to transform to martensite. Figures 3(a) and (b) show the changes in phase structure as a function of composition over ranges that encompass these alloys.

Martensite can be present in two different forms. The α' -form is the bcc magnetic form, while ϵ is a nonmagnetic, hcp (hexagonal close-packed) version. The formation of ϵ versus α' is related to the stacking fault energy of the alloy, which is given by (Ref 6):

$$Y^{300}_{SF} (\text{mJ m}^{-2}) = Y^{0}_{SF} + 1.59\text{Ni} - 1.34\text{Mn} + 0.06\text{Mn}^2 - 1.75\text{Cr} + 0.01\text{Cr}^2 + 15.21\text{Mo} - 5.59\text{Si} - 60.69(\text{C} + 1.2\text{N})^{1/2} + 26.27(\text{C} + 1.2\text{N})(\text{Cr} + \text{Mn} + \text{Mo})^{1/2} + 0.61[\text{Ni} \cdot (\text{Cr} + \text{Mn})]^{1/2} \quad (\text{Eq 2})$$

Epsilon martensite formation is favored in alloys of lower stacking fault energy. The fcc structures deform by slip between (111) planes. Viewed from these planes, the structure is a series of ABCABC atom arrangements. Slip between planes can result in an ABCA/CAB structure. This so-called stacking fault generates an hcp structure. With lower stacking fault energies, these are more readily

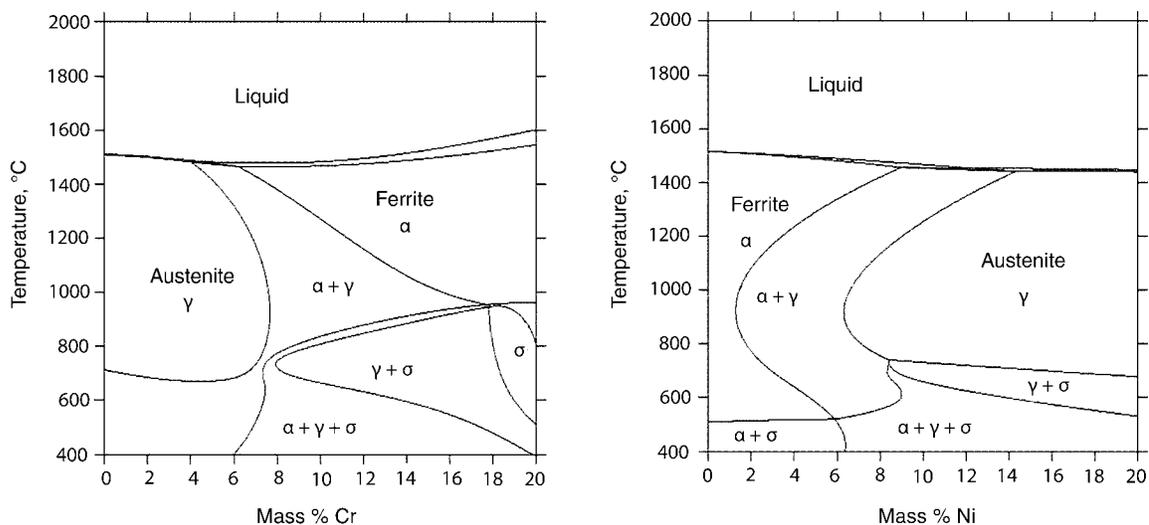


Fig. 3 (a) Iron-chromium phase diagram at 8% nickel; (b) iron-nickel phase diagram at 18% chromium

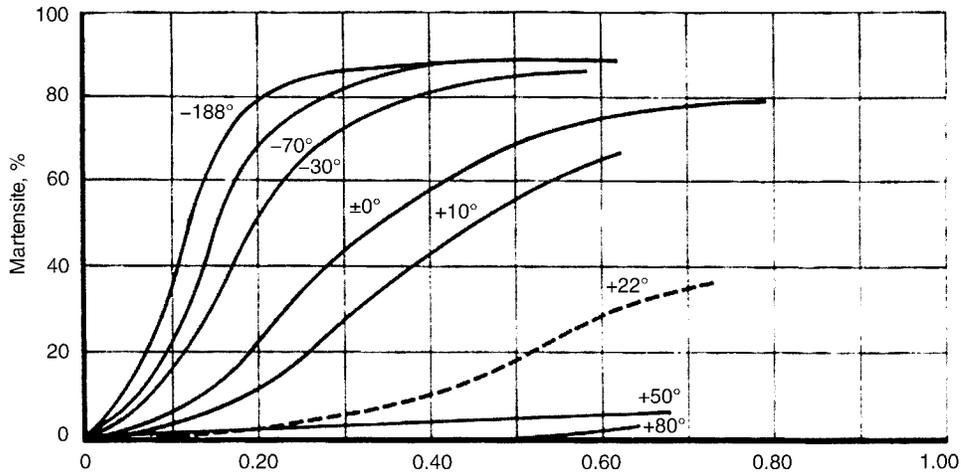


Fig. 4 Variation of martensite formation with temperature and true strain for 304. Source: Ref 7

formed, and ϵ predominates. The stacking fault can also be viewed as two partial dislocations with the material between them faulted. These partial dislocations, when generated in abundance, cannot readily slip past one another and thus pile up, increasing work-hardening rates.

As in carbon and alloy steels, the martensite transformation can take place simply by cooling, but in the lean austenitic alloys the temperatures are well below ambient. The more stable alloys do not transform even with cryogenic treatment. Figure 4 shows the variation of martensite formation with temperature and true strain for 304. Martensite formed in these alloys is quite stable and does not revert until heated well above the temperatures (Fig. 5) at which it was formed. The carbon levels of austenitic stainless steels are always relatively low, so strain-induced martensite is self-tempering and not brittle.

Martensite has been found to form in unstable austenite due to the electrochemically induced supersaturation by hydrogen (Ref 9). Under conditions of cathodic charging, superficial layers were found to transform to ϵ under conditions of intense hydrostatic compression. During subsequent outgassing, α' was found to form due to reversals in the stress state. Martensite thus formed is, of course, susceptible to hydrogen embrittlement.

Mechanical Properties. The tensile properties in the annealed state not surprisingly relate well to composition. The 0.2% yield strength

and tensile strength, respectively, are reported (Ref 10) to follow the equations:

$$\begin{aligned} YS(\text{MPa}) = & 15.4[4.4 + 23(\%C) + 32(\%N) \\ & + 0.24(\%Cr) + 0.94(\%Mo) \\ & + 1.3(\%Si) + 1.2(\%V) \\ & + 0.29(\%W) + 2.6(\%Nb) \\ & + 1.7(\%Ti) + 0.82(\%Al) \\ & + 0.16(\%Ferrite) \\ & + 0.46(d^{-1/2}) \end{aligned} \quad (\text{Eq 3})$$

$$\begin{aligned} TS(\text{MPa}) = & 15.4[29 + 35(\%C) + 55(\%N) \\ & + 2.4(\%Si) + 0.11(\%Ni) + 1.2(\%Mo) \\ & + 5.0(\%Nb) + 3.0(\%Ti) + 1.2(\%Al) \\ & + 0.14(\%Ferrite) + 0.82(d^{-1/2}) \end{aligned} \quad (\text{Eq 4})$$

In each case, d is the grain diameter in millimeters.

Another researcher (Ref 11) gave the relationships as:

$$\begin{aligned} YS(\text{MPa}) = & 120 + 210\sqrt{N + 0.02} + 2Mn + 2Cr \\ & + 14Mo + 10Cu + (6.15 - 0.054\delta)\delta \\ & + (7 + 35(N + 0.2))d^{-1/2} \end{aligned} \quad (\text{Eq 5})$$

$$\begin{aligned} TS = & 470 + 600(N + 0.2) \\ & + 14Mo + 1.5\delta + 8d^{-1/2} \end{aligned} \quad (\text{Eq 6})$$

Again, d is grain diameter in millimeters, and δ is percent ferrite. The claimed accuracy for the latter set of equations is 20 MPa and is said

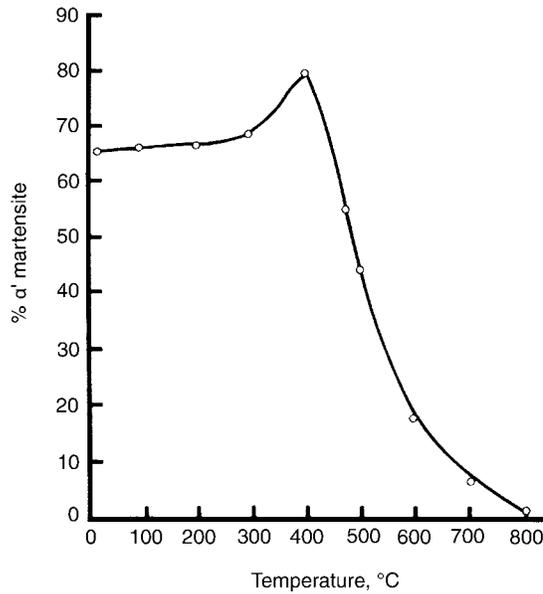


Fig. 5 Reversion of martensite formed by cold work. Source: Ref 8

to apply to both austenitics and duplex stainless steels, but clearly the tensile strength relationship must break down for leaner alloys, such as 301, in which tensile strength increases with decreasing alloy content because of the effect of increasing alloying causing less transformation to martensite, which inarguably produces higher tensile strengths in austenitic stainless steels. Equation 3 must also be favored over Eq 5 in that it accounts for carbon explicitly.

One other hardening mechanism is possible in austenitic stainless steels, and that is precipitation hardening. Most precipitation-hardening stainless steels are unstable austenite, which is transformed to martensite before the precipitation hardening takes place. One commercial alloy, A-286, is entirely austenitic and employs the precipitation within the austenite matrix of Ni_3 (titanium, aluminum) for strengthening. This is dealt with in a separate section.

Austenitic stainless steels do not have a clear yield point but can begin to deform at as little as 40% of the yield strength. As a rule of thumb, behavior at less than half the yield strength is considered fully elastic and stresses below two-thirds of the yield strength produce negligible plastic deformation. This quasi-elastic behavior is a consequence of the many active slip systems in the fcc structure. Even highly cold-worked material exhibits this phenomenon, although stress-relieving cold-worked material will cause dislocations to “lock in place” and

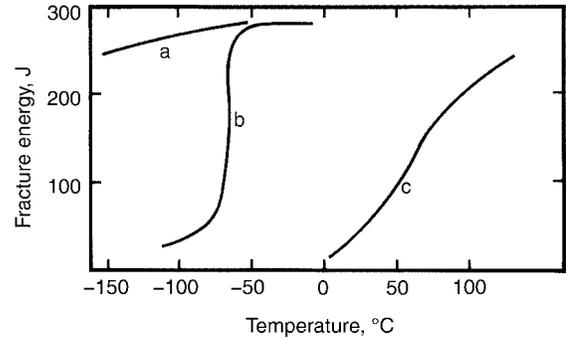


Fig. 6 Variation of impact strength with temperature for (a) austenitic, (b) duplex, and (c) ferritic stainless steels

form more stable dislocation arrays that break loose at a higher and distinct yield point.

The tensile properties of austenitic stainless steels with unstable austenite, that is, those with Md_{30} temperatures (Eq 1) near room temperature, are very strain rate dependent. This is simply due to the influence of adiabatic heating during testing increasing the stability of the austenite. Tests run under constant temperature conditions, either by slow strain rates or use of heat sinks, produce lower tensile strengths. Thus, reported tensile strengths should not be taken as an absolute value but a result that can be significantly changed by changes in testing procedure, even with accepted norms and standards.

Highly cold-worked austenitic stainless steels are often used for their robust mechanical properties. Few metallic materials can match the very high strengths they can achieve. Very lean 301 can be cold worked to yield strengths on the order of 2000 MPa because of its unstable austenite transforming to martensite. When cold worked to lower degrees, it can provide very high strength while keeping impressive ductility.

Austenitic stainless steels have exceptional toughness. The ambient temperature impact strength of austenitic stainless steels is quite high. This is not surprising in view of their high tensile strengths and high elongations. What is most remarkable is the absence of a transition temperature, which characterizes ferritic and martensitic materials. Figure 6 shows impact strength of the various stainless steel types versus temperature. This again is due to the multiplicity of slip systems in the fcc structure and the fact that they do not require thermal activation. This makes the austenitic stainless steels, especially the 200 series, the optimal cryogenic

material, surpassing the 9% nickel martensitic steels in cost, toughness, and, of course, corrosion resistance.

Precipitation of Carbides and Nitrides.

Carbon is normally considered as an undesirable impurity in austenitic stainless steel. While it stabilizes the austenite structure, it has a great thermodynamic affinity for chromium. Because of this affinity, chromium carbides, $M_{23}C_6$, form whenever carbon reaches levels of supersatura-

tion in austenite, and diffusion rates are sufficient for carbon and chromium to segregate into precipitates. The solubility of carbon in austenite is over 0.4% at solidification but decreases greatly with decreasing temperature. The solubility is given by (Ref 12):

$$\log(C \text{ ppm}) = 7771 - \frac{6272}{T (^{\circ}\text{K})} \quad (\text{Eq 7})$$

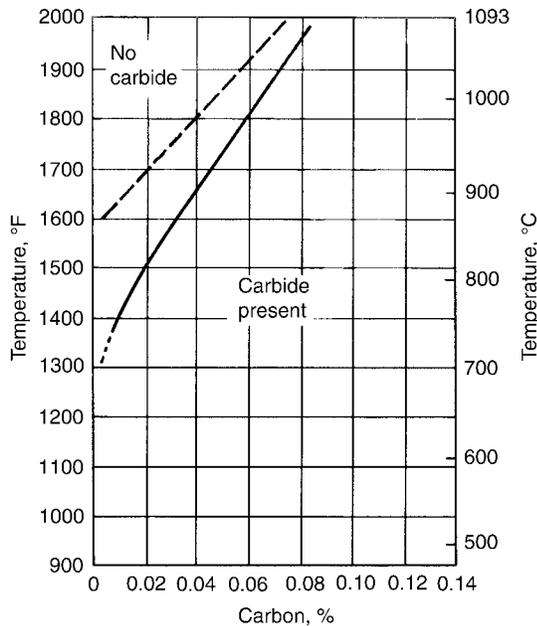


Fig. 7 Carbon solubility in 18-10 austenitic stainless. Source: Ref 13

The equilibrium diagram for carbon in a basic 18%Cr10%Ni alloy is shown in Fig. 7. At room temperature, very little carbon is soluble in austenite; even the 0.03% of L grades is mostly in a supersaturated solution. The absence of carbides in austenitic stainless is due to the slow diffusion of carbon and the even slower diffusion of chromium in austenite. At a carbon level of 0.06%, which is found in most 304, supersaturation is reached below about 850 °C. Below this temperature, supersaturation increases exponentially, while diffusion decreases exponentially. This results in precipitation rates that vary with temperature and carbon level as shown in Fig. 8. At these temperatures, grain boundary diffusion is much more rapid than bulk diffusion, and grain boundaries provide excellent nucleation sites, so precipitation occurs along grain boundaries. And, because carbon diffuses several orders of magnitude more rapidly than chromium, carbon diffuses to and combines with chromium essentially in situ, depleting the grain boundaries of chromium in solution.

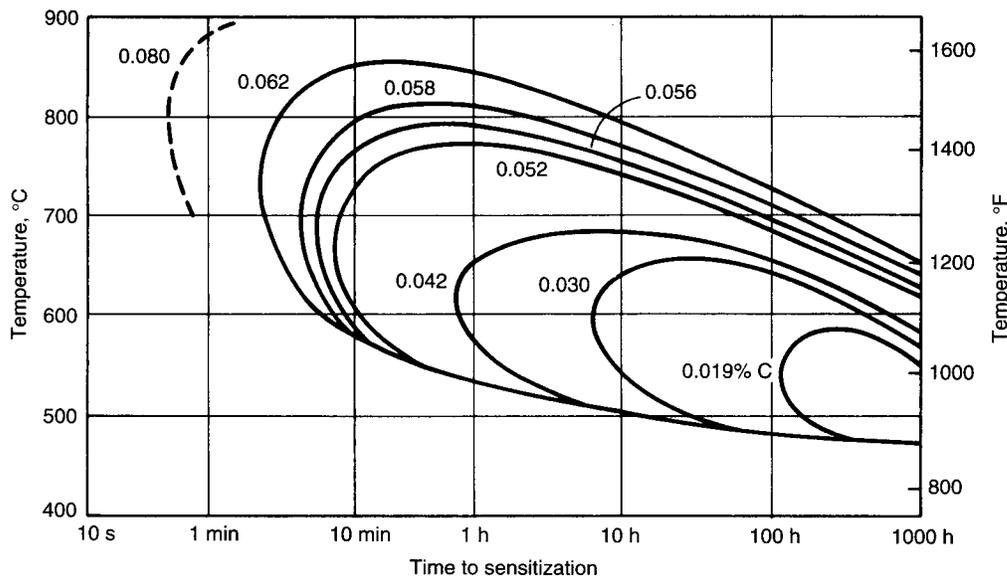


Fig. 8 The precipitation rates for $Cr_{23}C_6$ as a function of carbon content

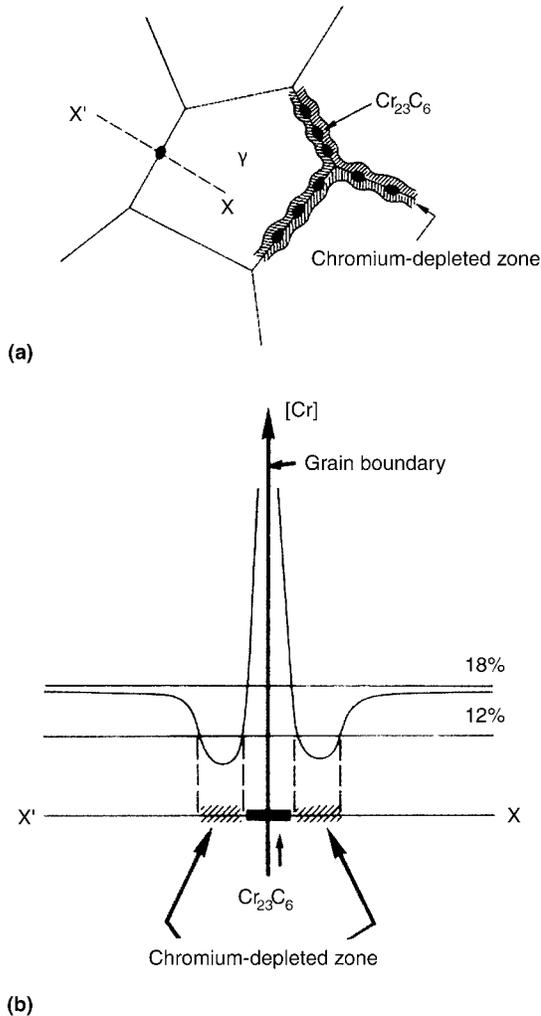


Fig. 9 Depletion of chromium from the austenite near grain boundaries due to chromium carbide precipitation. Source: Ref 14

Figure 9 shows that the local chromium depletion is such that the chromium level can become low enough that it has not even enough to be stainless and certainly much lower corrosion resistance than the surrounding area. This zone, because it is lower in chromium, also has very unstable austenite and is quite prone to martensite formation. Figure 10 shows how the locus of precipitation changes with time and temperature. Carbon relatively far from grain boundaries in the interior of grains remains in supersaturation until much longer times and much greater supersaturation since bulk diffusion is required for the nucleation and growth of these precipitates.

The key observation is that any solid-state precipitation of a chromium-rich precipitate will necessarily cause local chromium depletion and a resulting loss of corrosion resistance.

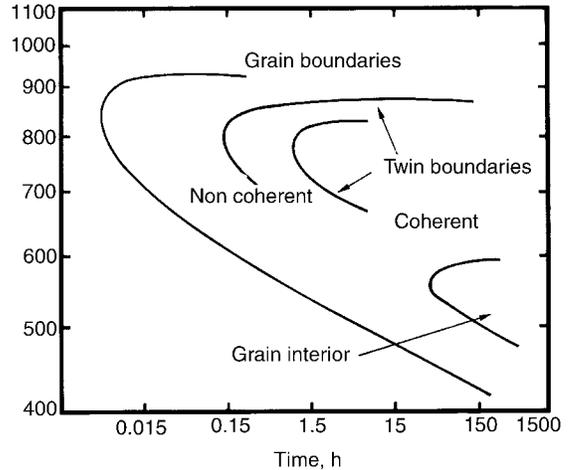


Fig. 10 Variation of carbide precipitation locus with time. Source: Ref 16

Much longer term heat treatment is required to eliminate these depleted zones by rehomogenization of slowly diffusing chromium than the short time required to form them. This is very evident for carbides, but also true for oxides. Underneath chromium-rich oxide scales is a layer depleted in chromium and lower in corrosion resistance. This is why not only scale from welding must be removed, but also the underlying chromium-depleted zone.

Other precipitation processes that give rise to chromium depletion are α and χ and the solid-state precipitation of oxides, nitrides, and sulfides. Chromium precipitates that form in the liquid alloy do not cause depletion of chromium locally because no chromium gradients are set up around them during precipitation as diffusion in the liquid is very rapid. Thus, primary carbides, oxides, and sulfides are not per se harmful to corrosion resistance. But, if the same compounds form and grow in the solid state, chromium depletion occurs (Ref 15).

Alloying elements can have a major influence on carbide precipitation by their influence on the solubility of carbon in austenite. Molybdenum and nickel accelerate the precipitation by diminishing the solubility of carbon. Chromium and nitrogen increase the solubility of carbon and thus retard and diminish precipitation. Nitrogen is especially useful in this regard (Fig. 11).

Increasing austenite grain size accelerates precipitation, as does cold work, especially in the interior of grains, where diffusion is enhanced by increased defect density.

Nitrogen is much more soluble than carbon and does not give rise to sensitization phenomena as does carbon even though Cr_2N can be a

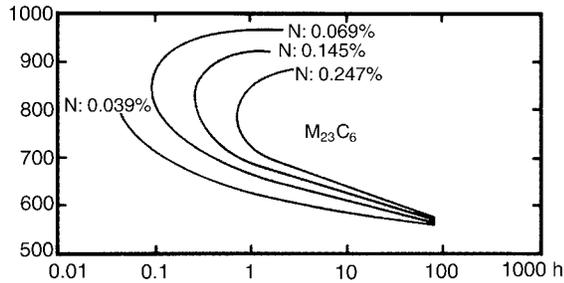


Fig. 11 Delay in carbide precipitation induced by nitrogen level. Source: Ref 17

stable phase when the solubility limit is exceeded. The solubility is over 0.15% in austenite, so its precipitation seldom has the possibility of occurring, but it does become an issue in ferritic stainless steels in this regard, for which solubility is much lower. Manganese and chromium increase the solubility of nitrogen in austenite.

Stabilization. Before carbon was easily lowered to harmless levels, it was found that adding more powerful carbide formers than chromium could preclude the precipitation of chromium carbides. Titanium and niobium are the most useful elements in this regard. They form carbides with solubility that follows the following equation type:

$$\log [M][X] = +A - \frac{H}{RT} \quad (\text{Eq 8})$$

For titanium carbide and niobium carbide, the respective solubilities are:

$$\log [\text{Ti}][\text{C}] = 2.97 - \frac{6780}{T} \quad (\text{Eq 9})$$

$$\log [\text{Nb}][\text{C}] = 4.55 - \frac{9350}{T} \quad (\text{Eq 10})$$

Oxides and sulfides are more energetically favorable than are carbides and nitrides of these metals. Thus, any additions made to form carbides must be sufficient to account for the prior formation of these compounds. Nitrogen also competes with carbon for available titanium or niobium. Thus, for successful gettering of all carbon, there must be sufficient titanium or niobium to combine stoichiometrically with all these species present.

This requires in rough terms that titanium exceed four times the carbon plus nitrogen, or that niobium exceed eight times, after accounting for the oxygen and sulfur. It would be a mistake

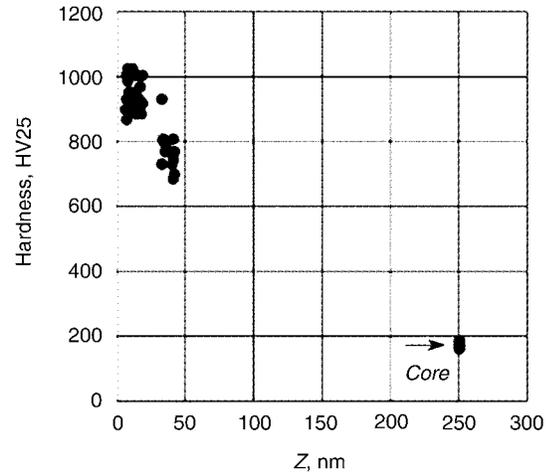


Fig. 12 Variation of hardness with depth and therefore carbon content in colossal supersaturation

to ignore the titanium-consuming capacity of oxygen and sulfur unless they have been minimized by refining, which can be done quite readily.

Even if sufficient titanium or niobium is present to combine with all carbon, kinetic considerations may result in that not occurring. High temperatures, such as encountered in welding, dissociate carbides. If quenched from this state, carbon can be free to form Cr_{23}C_6 if it is reheated to temperatures above 500°C .

Carbon has always been considered totally undesirable from a corrosion point of view because of its tendency to form chromium carbides. Recently, however, new processes have been developed to supersaturate carbon in austenite below the temperatures at which it has sufficient mobility to form carbides. This so-called colossally supersaturated austenite results in very high hardness (Fig. 12) and corrosion resistance over limited depths. From this, however, we can see that carbon, like nitrogen, is actually beneficial to corrosion resistance in solid solution, although this is not observed at normal concentrations. It is possible to see that if it could be kept in solution it would be appropriate to give it a factor of around 10 in the pitting resistance equivalent number (PREN) equation:

$$\text{PREN} = \% \text{Cr} + 3.3(\% \text{Mo}) + 30(\% \text{N}) + 10(\% \text{C}) \quad (\text{Eq 11})$$

This is consistent with the similar thermodynamic interaction coefficients that carbon and nitrogen share with regard to chromium.