MECH 423 Casting, Welding, Heat Treating and NDT

Time: _ _ W _ F 14:45 - 16:00

Credits: 3.5 Session: Fall

Heat Treatment

Lecture 6

Heat Treatment

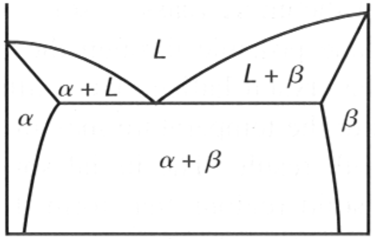
- This is a stupendously amazing concept that can produce some quite amazing changes in certain materials WITHOUT having to change component shape!
  - Softens cold-worked metals (annealing)
  - Strengthens some metals (quench & temper, ppt. harden)
  - Spheroidizes high carbon steels for easier machining.
  - Surfaces can be selectively hardened
  - Phase changes can be utilized for strengthening, shape memory affects etc.
- Requires understanding of PHASE DIAGRAMS!

Phase Diagrams (Review)

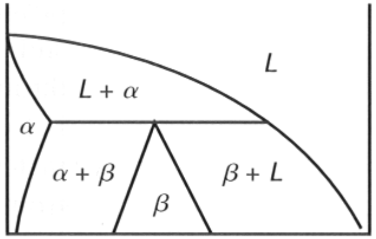
- **Phases** - “physically distinct, chemically homogeneous and mechanically separable portion of a substance”
- **Equilibrium phase diagrams** - variables are temperature, pressure, composition. E.g P-T, T-C.
- T-C phase diagrams determined by cooling curves. (**Liquidus, solidus, freezing range)**
- **Solubility-** solubility limits - non (insoluble), partial or complete solubility.
- Diagrams - give
  - phases present
  - composition of phases
  - amounts of phases present.

Three Phase Reactions

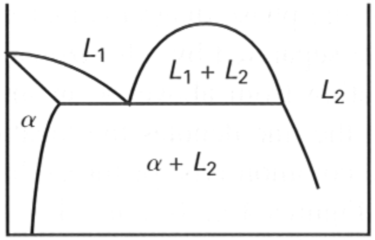
- E.g.
- Eutectic - Liquid → $\text{solid}_1 + \text{solid}_2$
- Peritectic - $L + S_1 \rightarrow S_2$
- Eutectoid - $S_1 \rightarrow S_2 + S_3$



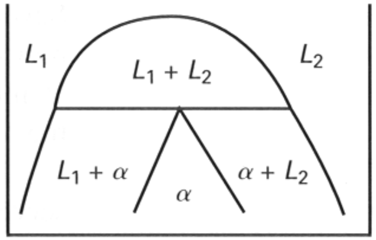
Eutectic
($L \rightarrow S_1 + S_2$)



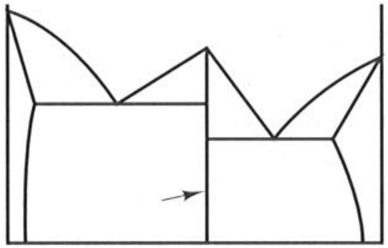
Peritectic
($L + S_1 \rightarrow S_2$)



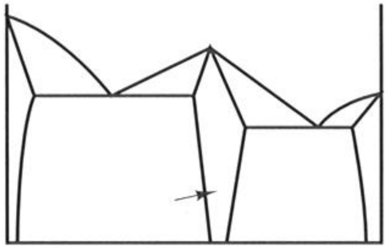
Monotectic
($L_1 \rightarrow S_1 + L_2$)



Syntectic
($L_1 + L_2 \rightarrow S_1$)



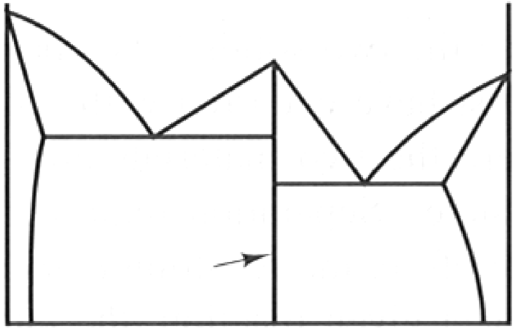
Stoichiometric
intermetallic compound



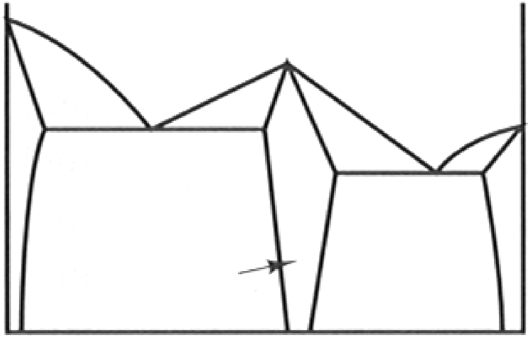
Non-stoichiometric
intermetallic compound

Compounds

- Sometimes formed at particular ratios of two metals or a metal and a non-metal.
- (two metals - Intermetallic compound - Fe_3Al).
- Fixed composition - Stoichiometric E.g. Fe_3C Iron carbide (cementite)
- These Compounds appear as Vertical lines on Phase Diagram.
- Some intermetallic compounds can exist over changes in composition - Non-stoichiometric
- Sometimes known as Intermediate phases
- Eg. Cu - Zn system, Mg - Pb system, Cu – Al system



Stoichiometric intermetallic compound



Non-stoichiometric intermetallic compound

Solid Phases in Fe-Fe_3C System

- **α - Ferrite**; Interstitial solid sol^n of C in BCC iron. Max solubility = 0.02%C at 723°C
- **γ - Austenite**; Interstitial solid Sol^n of C in FCC Iron; Max solubility = 2.08% at 1148°C (≈ 0.8 % at 723°C)

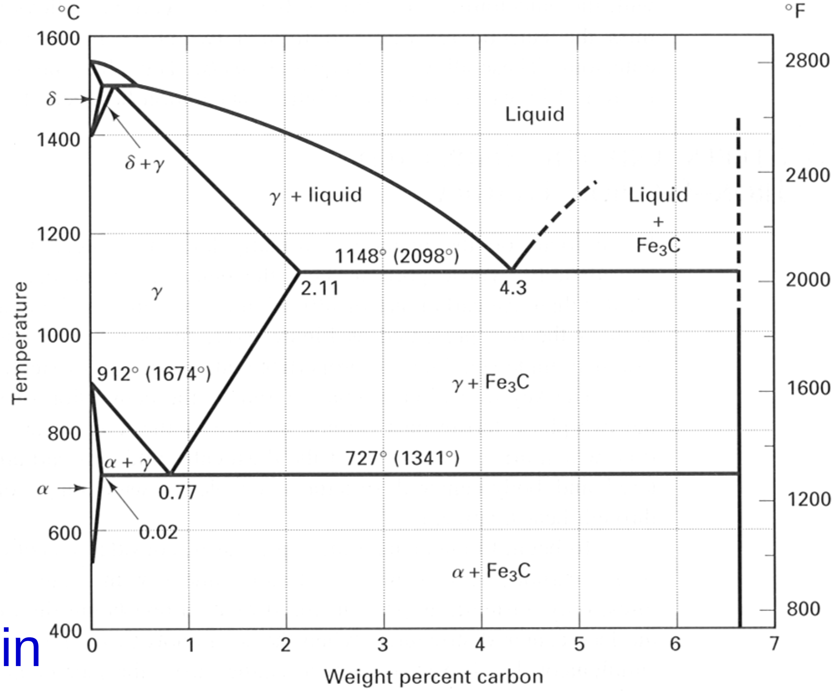


FIGURE 4.10 The iron–carbon equilibrium diagram: α, ferrite; γ, austenite; δ, δ–ferrite; Fe_3C, cementite.

- **δ - Ferrite** - Interstitial solid Sol^n of C in BCC iron. (lattice constant "a" is larger than for α - ferrite). Max solubility = 0.09%C at 1485°C.
- **Fe_3C - Cementite** - Iron Carbide (Intermetallic Compound). (3 atoms Fe to 1 atom C) or 93.3 wt% Fe + 6.67 wt% C (Hard + Brittle)

Invariant Points in Fe-Fe_3C System

Lever Rule

- **Peritectic** @ 1495°C

  Liq + δ → γ

  (0.53%C) (0.09%) (0.17%)

- **Eutectic** @ 1148°C

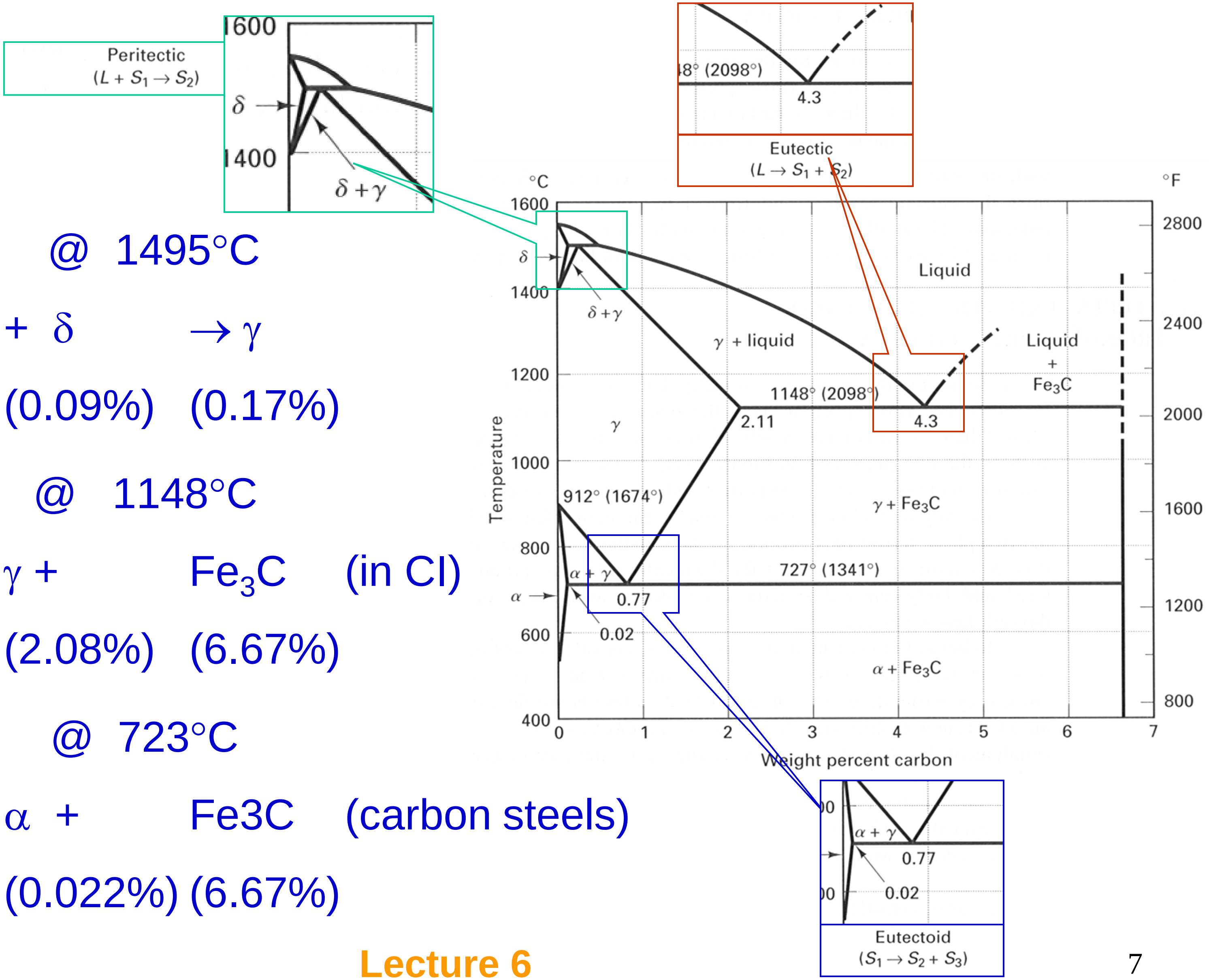
  Liq → γ + Fe_3C (in CI)

  (4.3%) (2.08%) (6.67%)

- **Eutectoid** @ 723°C

  γ → α + Fe3C (carbon steels)

  (0.77%) (0.022%) (6.67%)

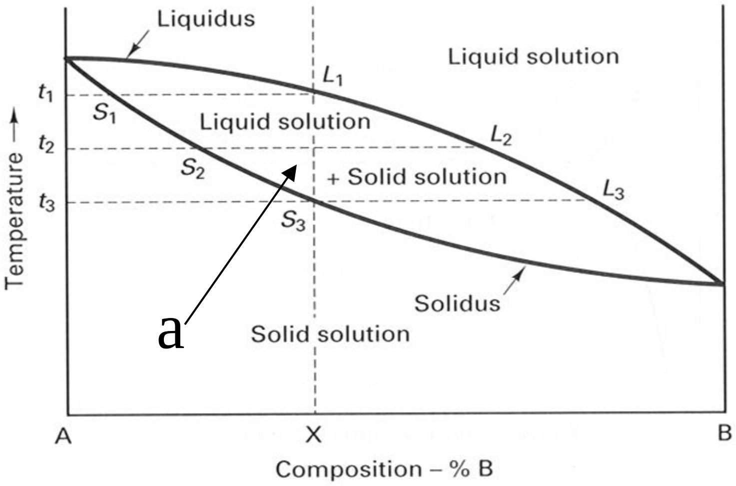


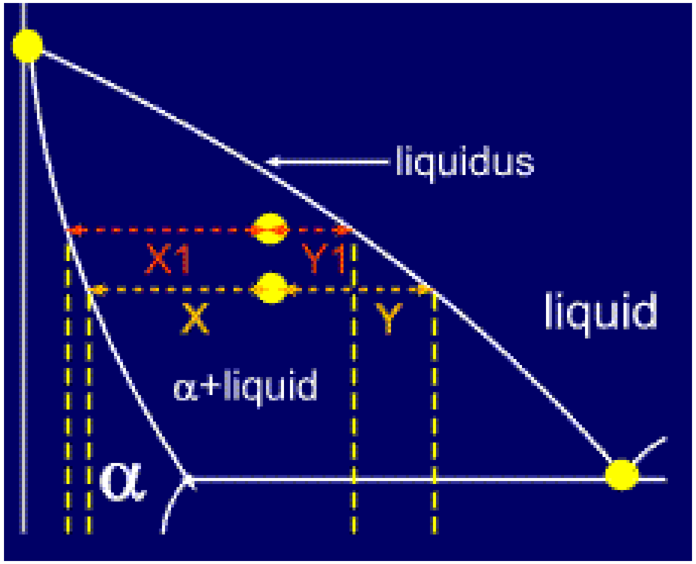
Lever Rule

- > t_1 the alloy X is liquid and < t_3 it is solid
- At t_1 the tie line runs from S_1 to L_1 and is to left of composition X
- At t_3 the tie line runs from S_3 to L_3 and is to right of composition X
- At t_2, it can be proved that the line to the left of point a gives the proportion of liquid and the line to the right gives the proportion of solid
- At t_2 the proportion can be calculated by

$$\frac{a - S_2}{L_2 - S_2} X100\%$$

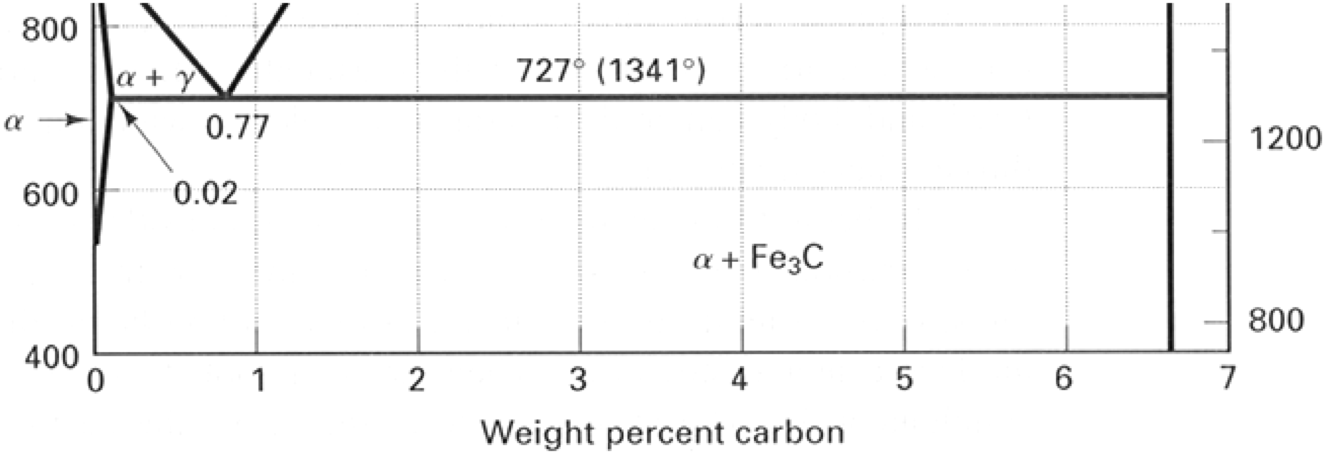
FIGURE 4.8 Equilibrium diagram showing the changes that occur during the cooling of alloy X.





The Eutectoid Reaction

- This particular combination of phases formed through this reaction is known as **Pearlite** (particular mixture of these two phases (α+Fe_3C) formed through the eutectoid reaction)
- As Eutectoid Reaction - (like eutectic) lamellar structure formed)
- So **Pearlite** (α + Fe_3C) formed by the Eutectoid Reaction is



$$\gamma(0.77\ wt\ \%C) \underset{heating}{\overset{cooling}{\rightleftharpoons}} \alpha(0.02\ \ wt\ \%C) + Fe_3C(6.7wt\ \%C)$$

The Eutectoid Reaction

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- So **Pearlite** (α + Fe_3C) formed by the Eutectoid Reaction is

$$\gamma(0.77\ wt\ \%C) \underset{heating}{\overset{cooling}{\rightleftharpoons}} \alpha(0.02\ \ wt\ \%C) + Fe_3C(6.7 wt\ \%C)$$

$$\%\alpha = \frac{6.67-0.77}{6.67-0.0218} x100 = 88.7\%$$

$$\%Fe_3C = \frac{0.77-0.0218}{6.67-0.0218} x100 = 11.3\%$$

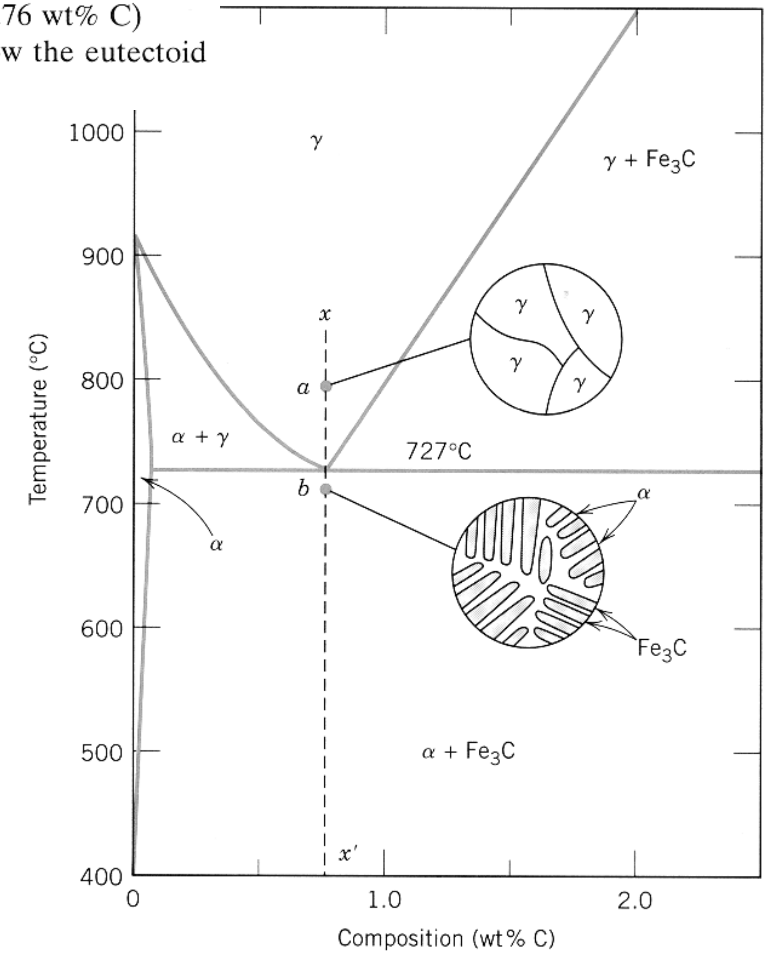
Microstructure Development in Fe-C

- 0.77wt%c **Eutectoid** Steel
- < 0.77wt%c **Hypo**eutectoid
- >0.77wt%c **Hyper**eutectoid

- Slow Cooling of Eutectoid Steel
- A eutectoid steel heated into the γ region (austenitized) and then slow cooled below 727°C will go through the eutectoid reaction:

$$\gamma \rightarrow \alpha + \mathbf{Fe_3C} \quad \text{(form 100\% pearlite)}$$

FIGURE 9.24 Schematic representations of the microstructures for an iron–carbon alloy of eutectoid composition (0.76 wt% C) above and below the eutectoid temperature.



Lecture 6

Microstructure Development in Fe-C

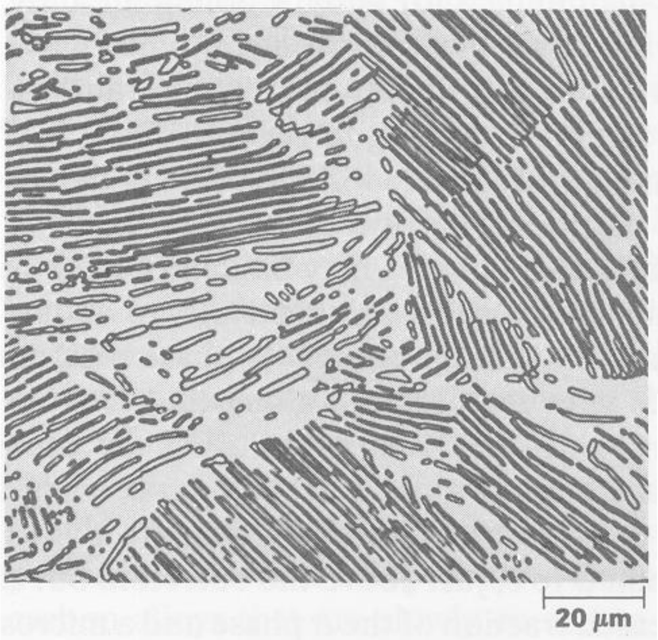


FIGURE 9.25 Photomicrograph of a eutectoid steel showing the pearlite microstructure consisting of alternating layers of α ferrite (the light phase) and Fe_3C (thin layers most of which appear dark). 500×.

- Pearlite is lamellar structure with alternate layers of Ferrite and cementite

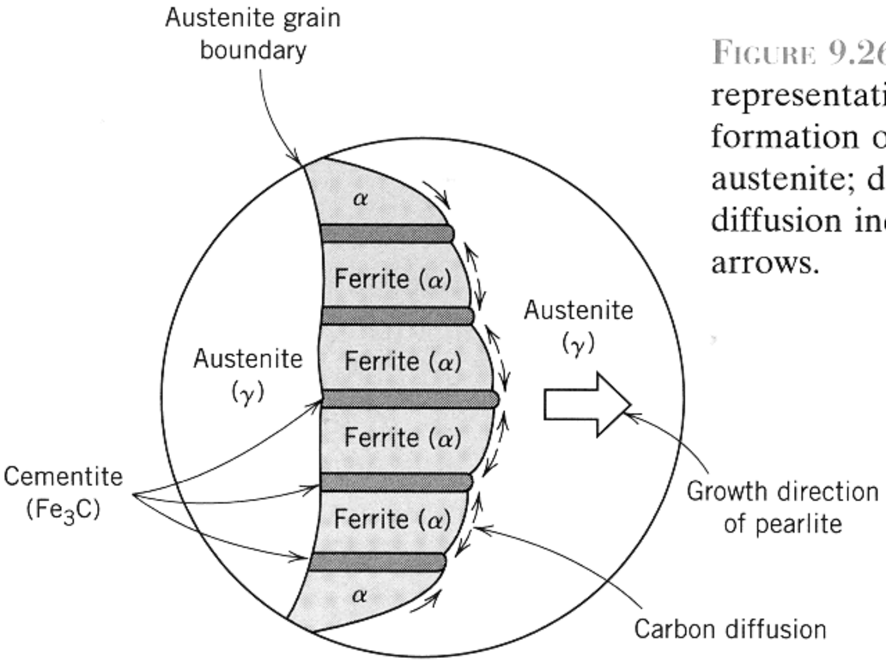


FIGURE 9.26 Schematic representation of the formation of pearlite from austenite; direction of carbon diffusion indicated by arrows.

Hypoeutectoid Steels

(< 0.77%C) eg 0.4%C steel

- Have to heat to higher T to get 100% γ (austenite) » 900°C

  Slow Cooled:

  $\gamma \rightarrow$ 800°C $\rightarrow \gamma + \alpha'$ (Proeutectoid/Primary Ferrite)

  $\gamma + \alpha' \rightarrow$ 712°C $\rightarrow \alpha' + \alpha + Fe_3C$ (Pearlite)
- Amounts of each:
- In this case because we want to distinguish between α and α' calculate just above transformation temp to get α'.
- Final grain consists proeutectoid α, formed before cooling down; and alternate layers of eutectoid α, and Fe_3C forming the pearlite structure

Hypoeutectoid Steels

- Increasing amounts of α while cooling - calculated using lever rule

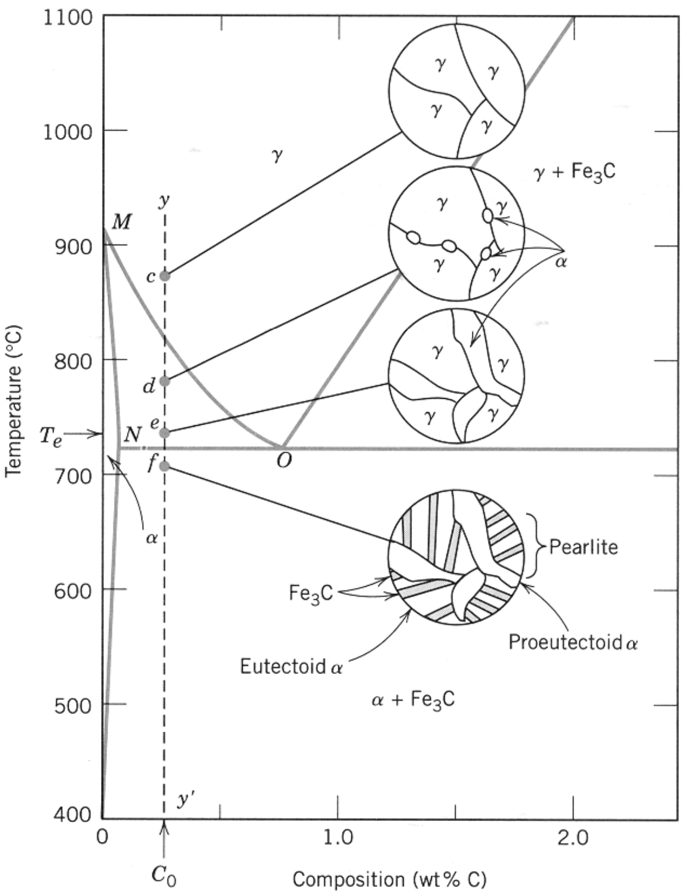
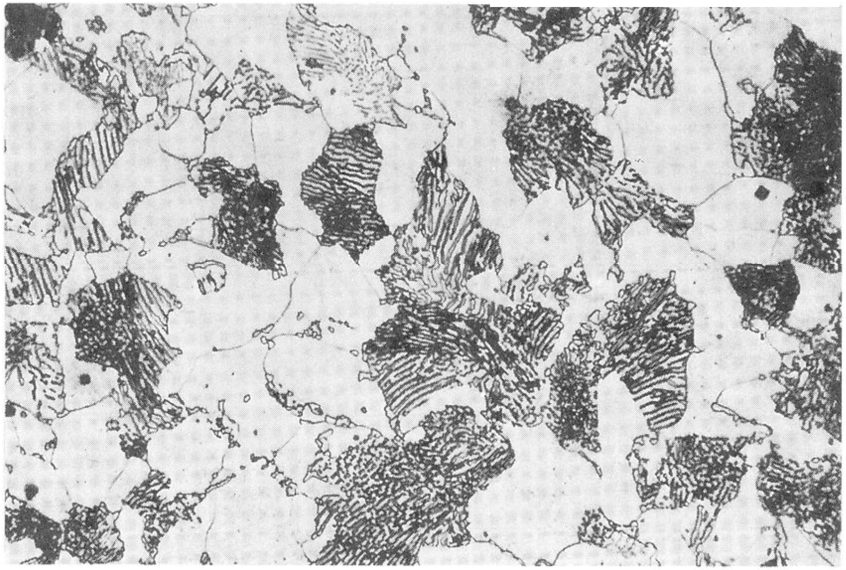


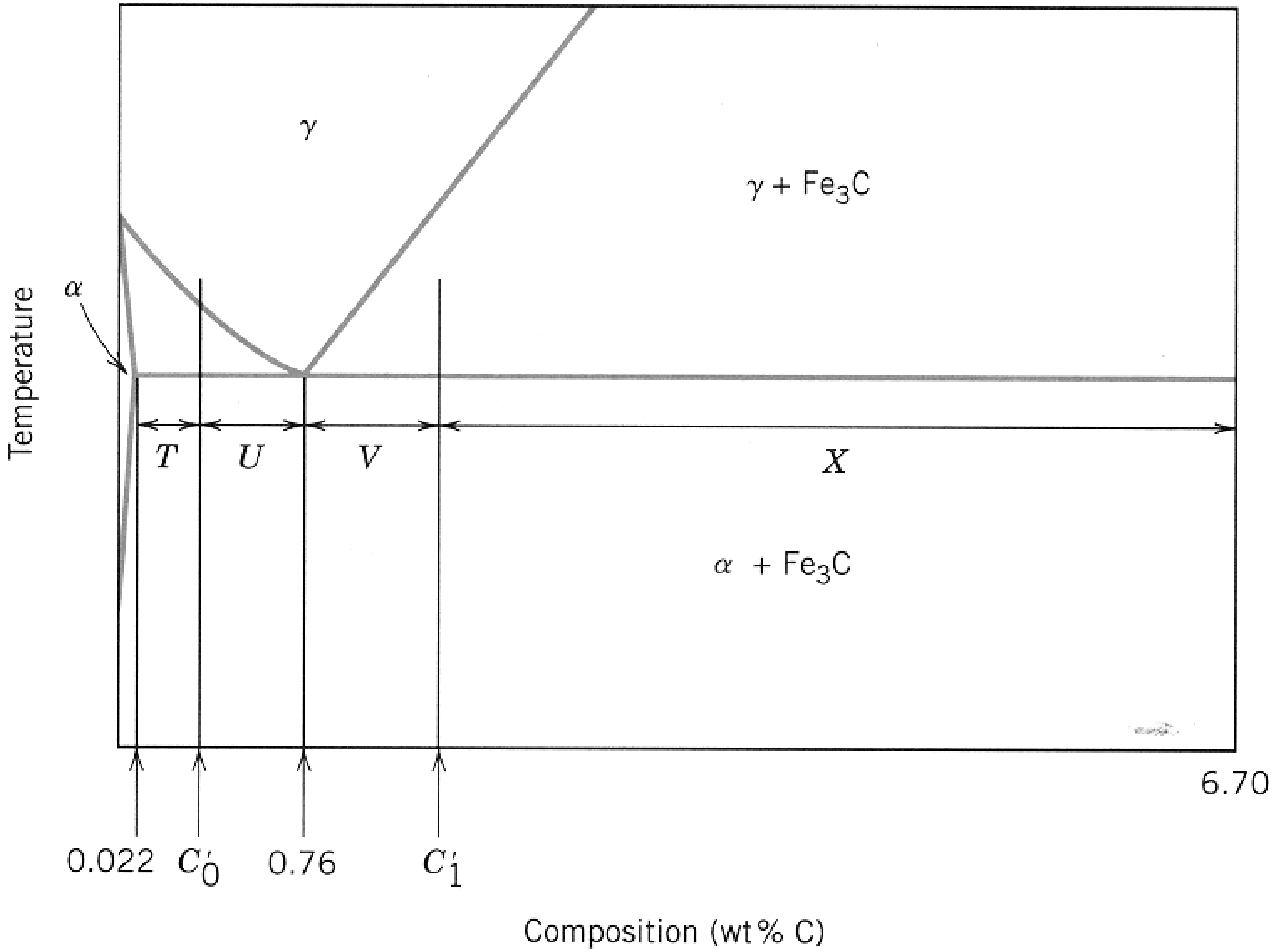
FIGURE 9.27 Schematic representations of the microstructures for an iron–carbon alloy of hypoeutectoid composition C_0 (containing less than 0.76 wt% C) as it is cooled from within the austenite phase region to below the eutectoid temperature.

FIGURE 9.28 Photomicrograph of a 0.38 wt% C steel having a microstructure consisting of pearlite and proeutectoid ferrite. 635×.



Hypoeutectoid Steels

FIGURE 9.29 A portion of the Fe–Fe_3C phase diagram used in computations for relative amounts of proeutectoid and pearlite microconstituents for hypoeutectoid (C_0') and hypereutectoid (C_1') compositions.



Hypereutectoid Steels

(> 0.77%C) eg 1.2%C steel

- Heat to ≈ 950^{o}C to Austenitize (slow cool)

  $\gamma \quad \rightarrow \quad \gamma \quad + \quad (Fe_3C)'$ (Proeutectoid/ Primary Fe_3C)

  1.2%C 1.1% 6.67%

  Then at 727^{o}C:

  $\gamma \quad + \quad Fe_3C \rightarrow \alpha \; + \quad Fe_3C \quad + \quad Fe_3C'$ pearlite

- So below Eutectoid temp (723^{o}C) there is Fe_3C and pearlite.
- Proeutectoid Cementite usually appears as white band along grain boundaries and Pearlite is lamallae.

Hypereutectoid Steels

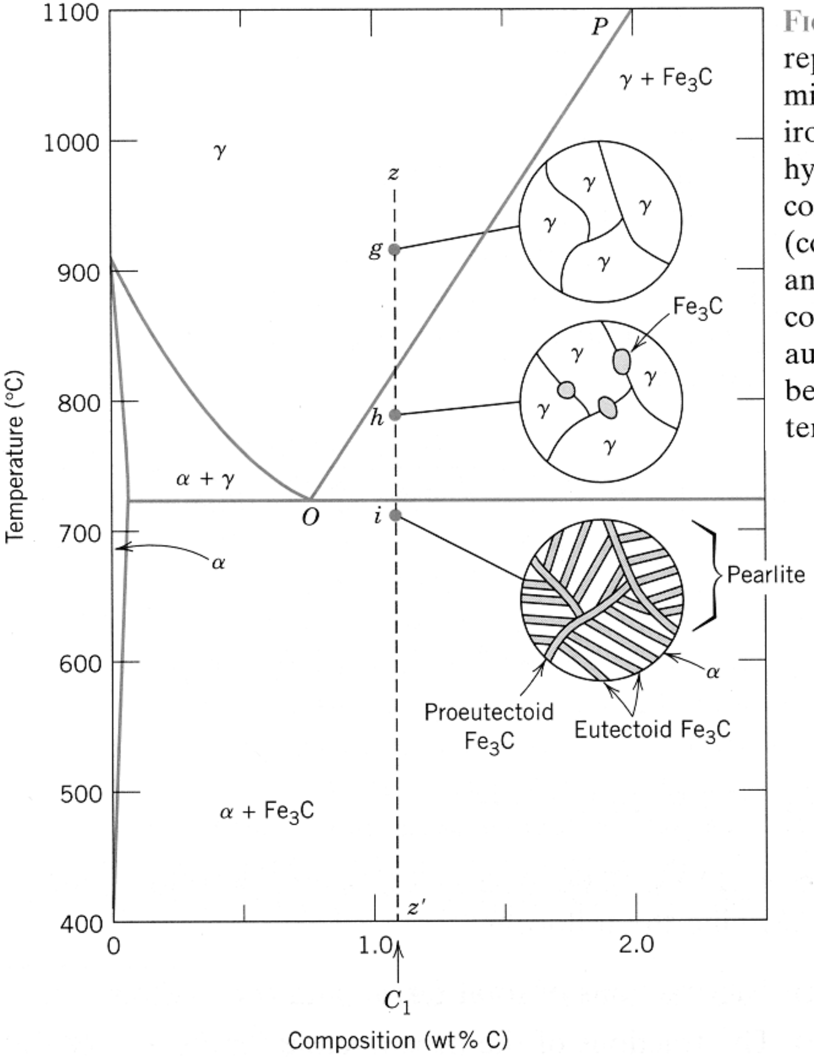


FIGURE 9.30 Schematic representations of the microstructures for an iron–carbon alloy of hypereutectoid composition C_1 (containing between 0.76 and 2.14 wt% C), as it is cooled from within the austenite phase region to below the eutectoid temperature.

FIGURE 9.31 Photomicrograph of a 1.4 wt% C steel having a microstructure consisting of a white proeutectoid cementite network surrounding the pearlite colonies. 1000×.

Hypoeutectoid Steels

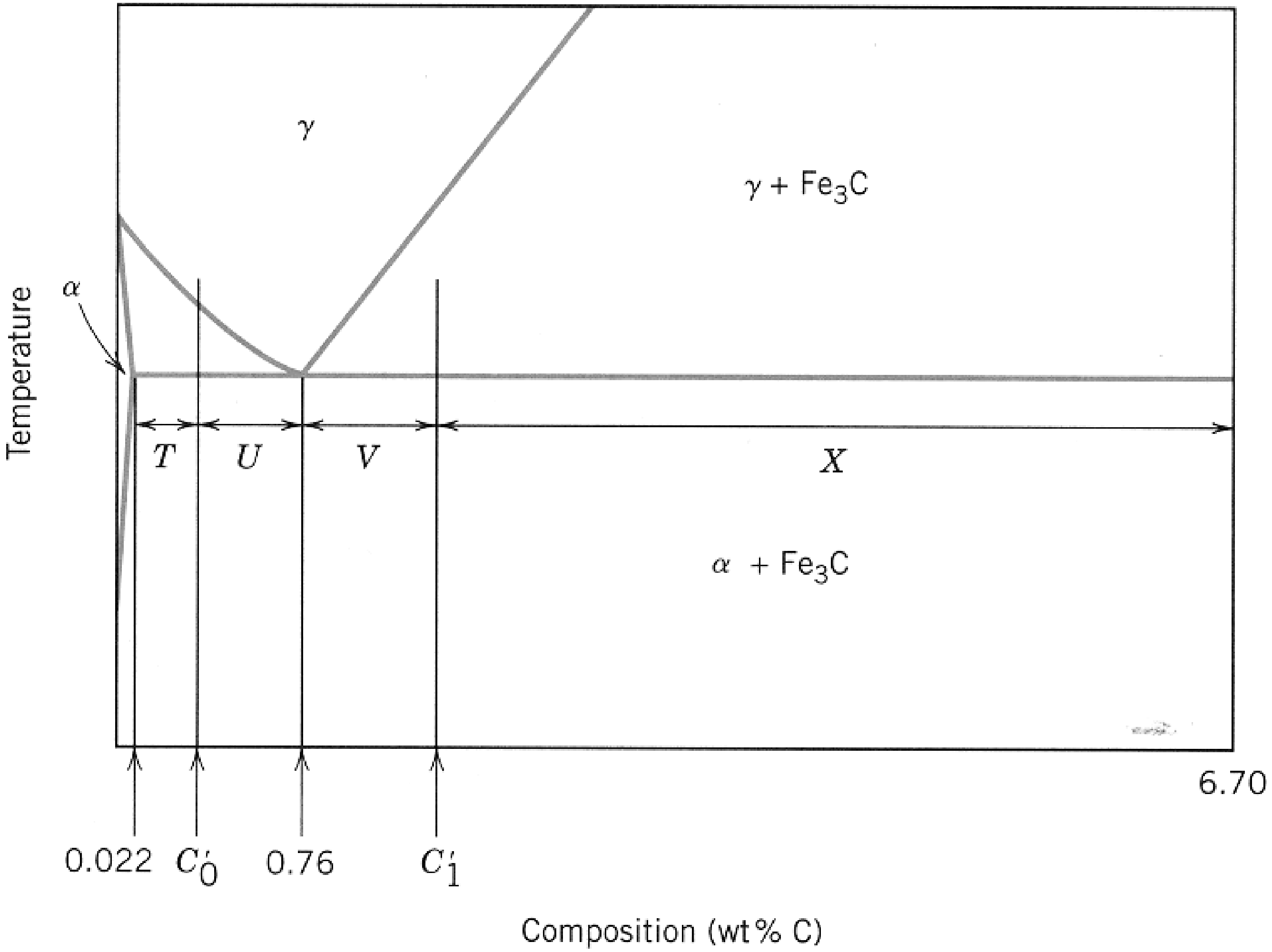


FIGURE 9.29 A portion of the Fe–Fe_3C phase diagram used in computations for relative amounts of proeutectoid and pearlite microconstituents for hypoeutectoid (C'_0) and hypereutectoid (C'_1) compositions.

Hypoeutectoid Steels

For a 99.65%Fe and 0.35C alloy at temp below eutectoid calculate the fractions of

total ferrite and cementite phases

$$\%W_{\alpha} = \frac{6.7 - 0.35}{6.67 - 0.02} x100 = 95\% \qquad \%W_{Fe3C} = \frac{0.35 - 0.02}{6.67 - 0.02} x100 = 5\%$$

Proeutectoid ferrite and pearlite phases

$$\%W_{p} = \frac{0.35 - .022}{0.76 - 0.022} x100 = 44\% \qquad \%W_{\alpha'} = \frac{0.76 - 0.35}{0.76 - 0.022} x100 = 56\%$$

eutectoid ferrite is total ferrite without proeutectoid ferrite

$$\%W_{\alpha e} = \%W_{\alpha} - \%W_{\alpha'} \qquad \%W_{\alpha e} = 95 - 56 = 39\%$$

Influence of Other Alloying Elements

- Elements like Titatnium, Chromium have significant effect on the eutectoid temperature and eutectoid composition (%C) of Iron
- In addition they also influence formation of pearlite with varying fractions of proeutectoid phase
- Generally steels are alloyed for different reasons like corrosion resistance etc.

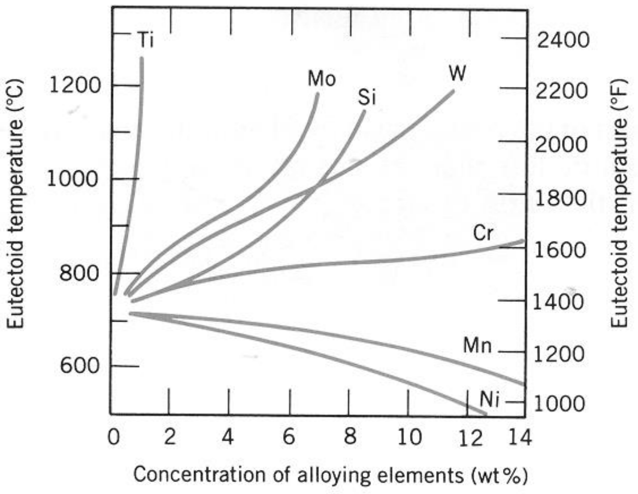


FIGURE 9.32 The dependence of eutectoid temperature on alloy concentration for several alloying elements in steel.

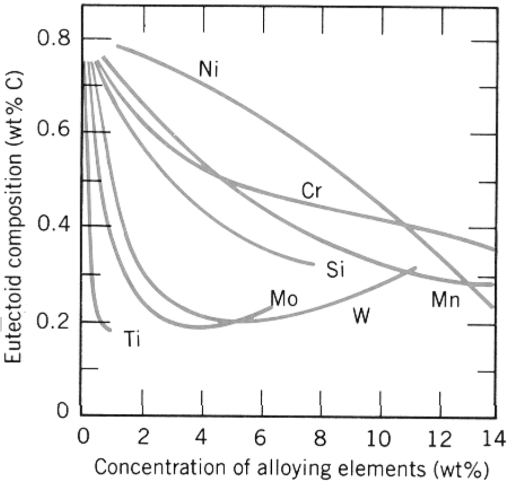


FIGURE 9.33 The dependence of eutectoid composition (wt% C) on alloy concentration for several alloying elements in steel.

Controlling Eutectoid Reaction

- Amount: more eutectoid (hard) by increasing %C initially.
- Austenite grain size: reduce γ grain size means smaller pearlite colonies ® higher strength
- Cooling rate: Increased cooling rate means finer lamallae ® higher strength
- OTHER STRUCTURES
- These can be produced by different types of cooling and thermal treatments. Eg. Bainite, Martensite

Processing Heat Treatments

- For Steels:
- Annealing - generally describes a heat treatment which will soften metal, or remove certain affects.
- Three stages of annealing
  - Heating to the desired temperature
  - Holding or “soaking” at that temperature
  - Cooling, usually to room temperature

Purpose of Annealing

- Relieve Internal Stresses
  - Internal stresses can build up in metal as a result of processing.
    - Stresses may be caused by previous processing operations such as welding, cold working, casting, forging, or machining.
  - If internal stresses are allowed to remain in a metal, the part may eventually distort or crack.
  - Annealing helps relieve internal stresses and reduce the chances for distortion and cracking.

Purpose of Annealing

- Increasing Softness, Machinability, and Formability
  - A softer and more ductile material is easier to machine in the machine shop.
  - An annealed part will respond better to forming operations.
- Refinement of Grain Structures
  - After some types of metalworking (particularly cold working), the crystal structures are elongated.
  - Annealing can change the shape of the grains back to the desired form.

Annealing

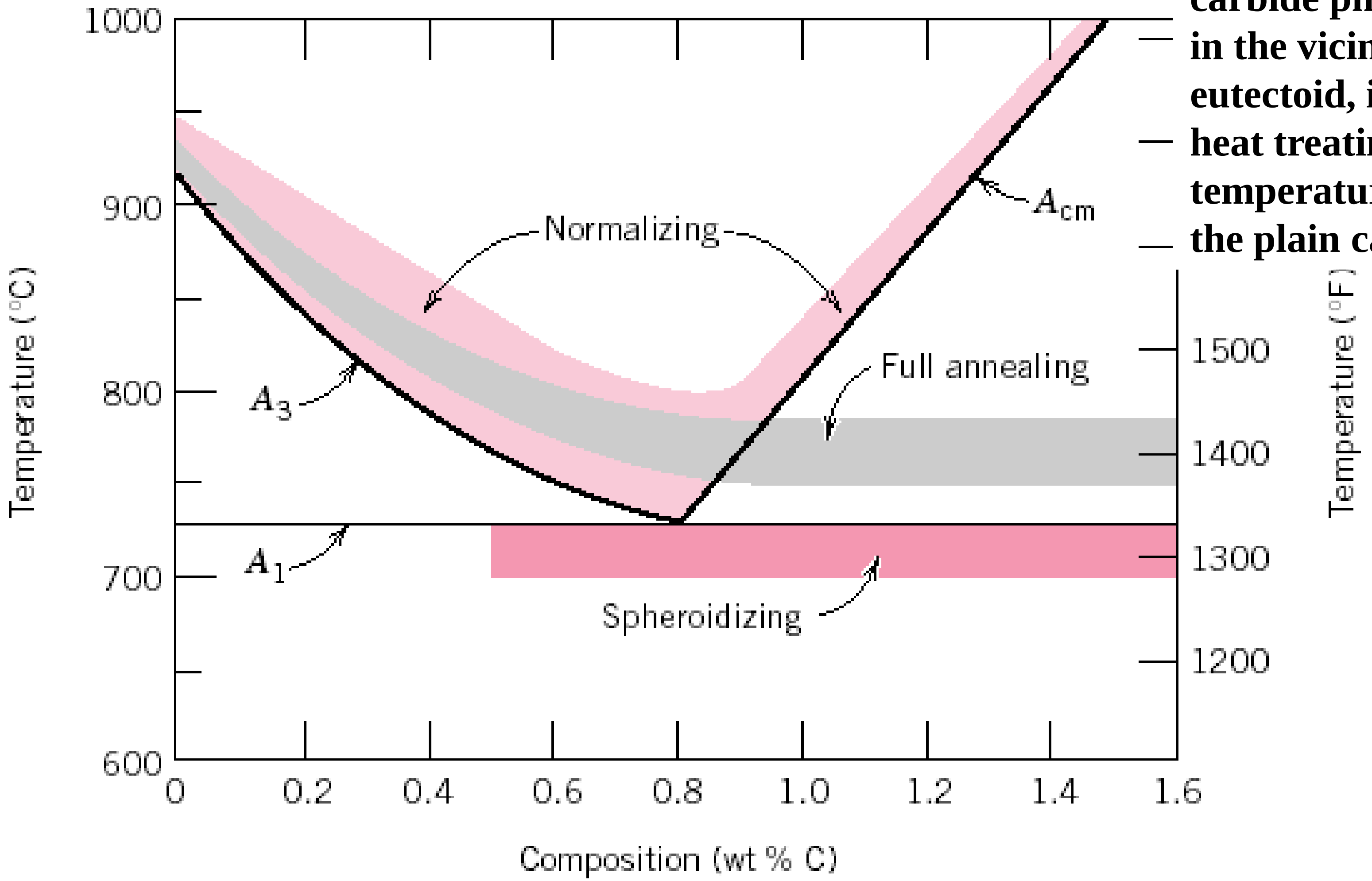


FIG. 11.9 **The iron-iron carbide phase diagram in the vicinity of the eutectoid, indicating heat treating temperature ranges for the plain carbon steels.**

Full annealing

- For hypoeutectoid steels - Heat into austenite region (A_3 + 50°C). Sufficient time for full austenitization, then cooled slowly in the furnace (≈ 20°C per hour) to less than ≈ 690°C then air-cooled. Gives a coarse pearlitic (and ferrite) structure - soft and ductile.
- For hypereutectoid steels - similar except heat into austenite + cementite region (A_1 + 50°C). Gives a coarse pearlitic (and spheroidal cementite) structure.
- Time and energy consuming (uniform properties throughout structure).

Process annealing

- A heat treatment used to negate the effects of cold work, i.e., to soften and increase the ductility of a previously strain-hardened metal
- In process annealing, parts are not as completely softened as they are in full annealing, but the time required is considerably lessened.
- Process annealing is frequently used as an intermediate heat-treating step during the manufacture of a part.

  A part that is stretched considerably during manufacture may be sent to the annealing oven three or four times before all of the stretching is completed.

Alteration of Grain Structure as a Result of Plastic Deformation

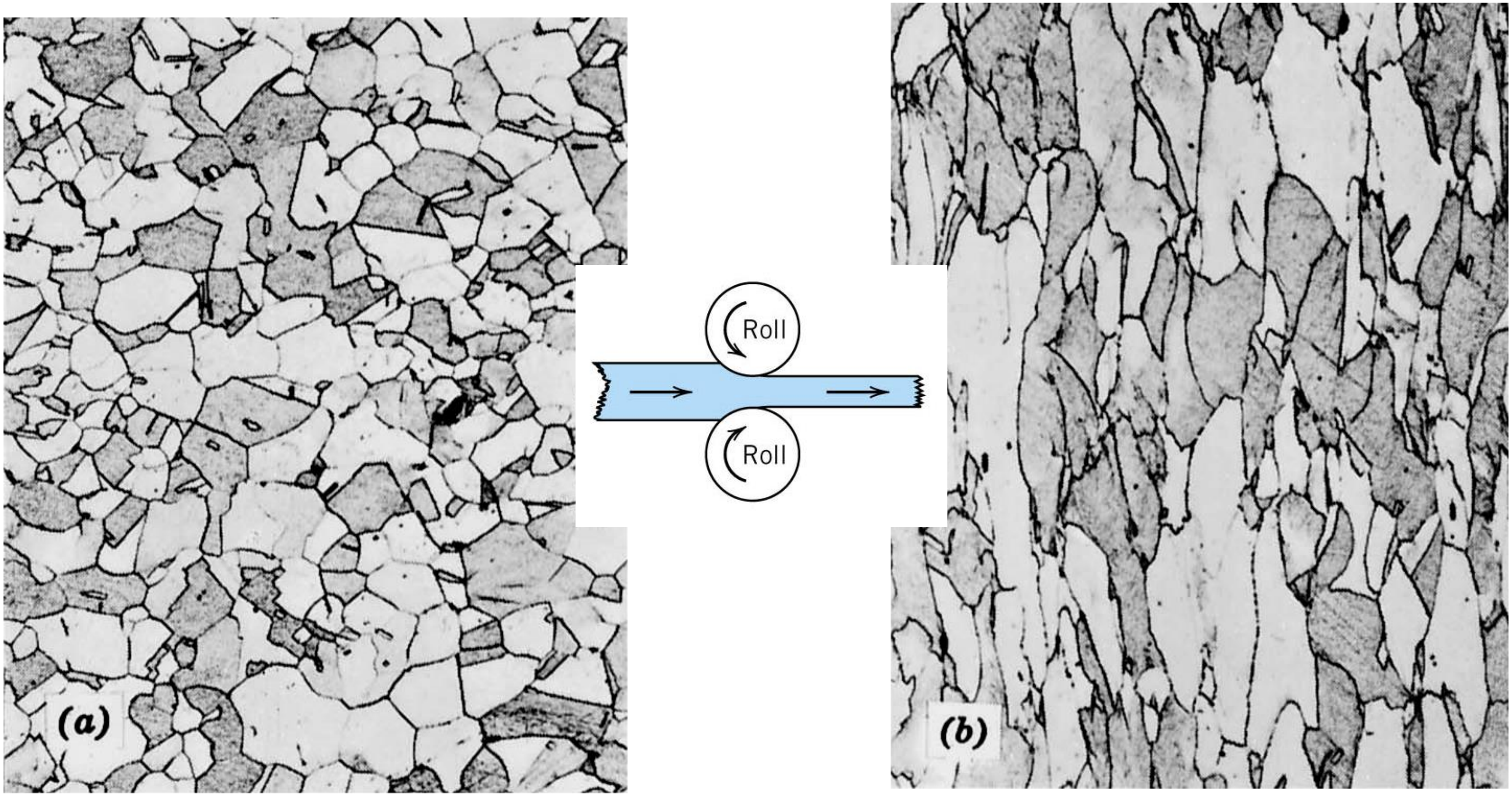


FIG. 7.11 **Alteration of the grain structure of a polycrystalline metal as a result of plastic deformation. (*a*) Before deformation the grains are equiaxed. (*b*) The deformation has produced elongated grains.**

Process annealing

- Primarily used to restore ductility to low carbon (0.25%) steels (c.f. recrystallization) during cold-working operations.
- steels are heated to below A_1 (10-20^{o}C below) typically 600-700^{o}C and held long enough for ferrite recrystallisation and then cooled.
- Changes ferrite morphology and structure but does not significantly affect carbides or induce phase changes.
- Lower temperatures so cheaper and less scaling.

Process annealing

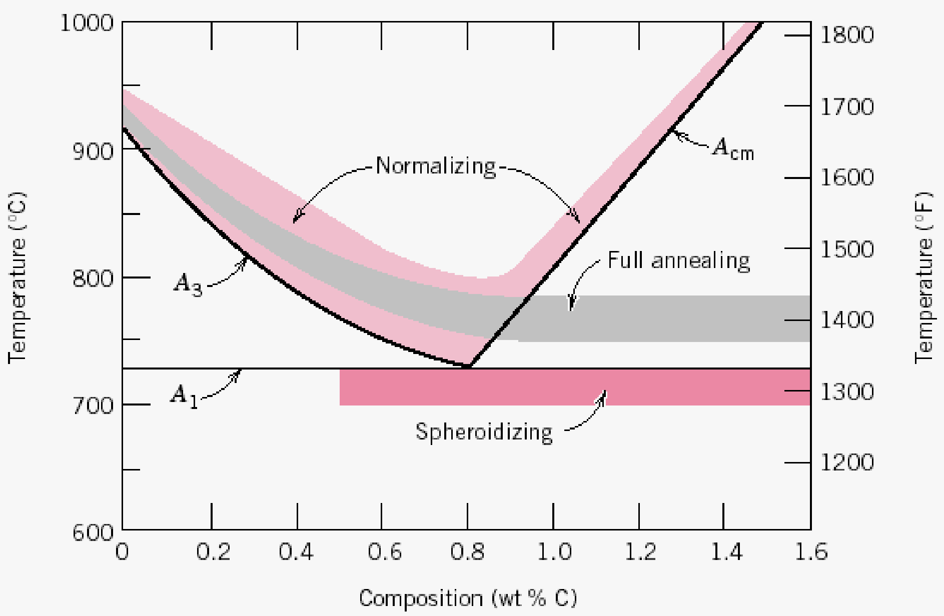
- Recovery and recrystallization processes are allowed to occur
- 1. Recovery
- Some of the stored internal strain energy is relieved by virtue of dislocation motion, as a result of enhanced atomic diffusion at the elevated temperature.
- 2. Recrystallization
- Recrystallization is the formation of a new set of strain free and equiaxed grains that have low dislocation densities and are characteristic of the precold-worked condition.
- Ordinarily a fine-grained microstructure is desired; the heat treatment is terminated before appreciable grain growth has occurred.

Stress Relief Annealing

- Can be used to relieve residual stresses in large steel castings, welded assemblies, cold-formed products.
- Heated to below A_1, typically 500-600^{o}C)and then cooled.

Normalizing

- The name "normalizing" comes from the original intended purpose of the process — to return steel to the "normal" condition it was in before it was altered by cold working or other processing.



- Heating the alloy to 55 to 85°C above the A_3 or A_{cm} and holding for sufficient time so that the alloy completely transforms to austenite, followed by air cooling

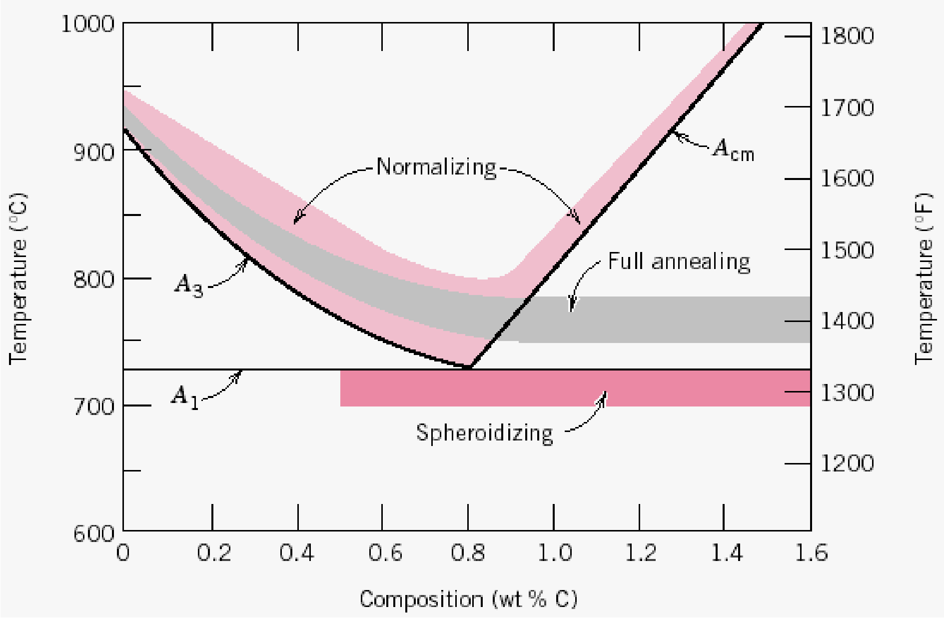
FIG. 11.9 **The iron-iron carbide phase diagram in the vicinity of the eutectoid, indicating heat treating temperature ranges for the plain carbon steels.**

Normalizing

- To refine grains and produce a more uniform and desirable size distribution for steels that are plastically deformed

- Normalizing does not soften the material as much as full annealing

- The cooling process does not leave the material as ductile or as internally stress-free.



  - A normalized part will usually be a little stronger, harder, and more brittle than a full-annealed part.

Spheroidizing

- Used to soften higher carbon (>0.6%) steels having high cementite contents; pearlite and primary Fe_3C.
- Want to make the Fe3C forms spheroids; this softens & toughens the higher C steels. (useful for machining and cold-forming operations).
  - Held just below Eutectoid temp (A1) for a long time; slow cooled.
  - Prolonged cycling above and below A1; slow cooled.
  - For tool steels; Heat to 750-800°C, hold for several hours and slow cool.
- NOTE: Increase hardness of these materials by subsequent normalizing, or Q&T.

Spheroidizing

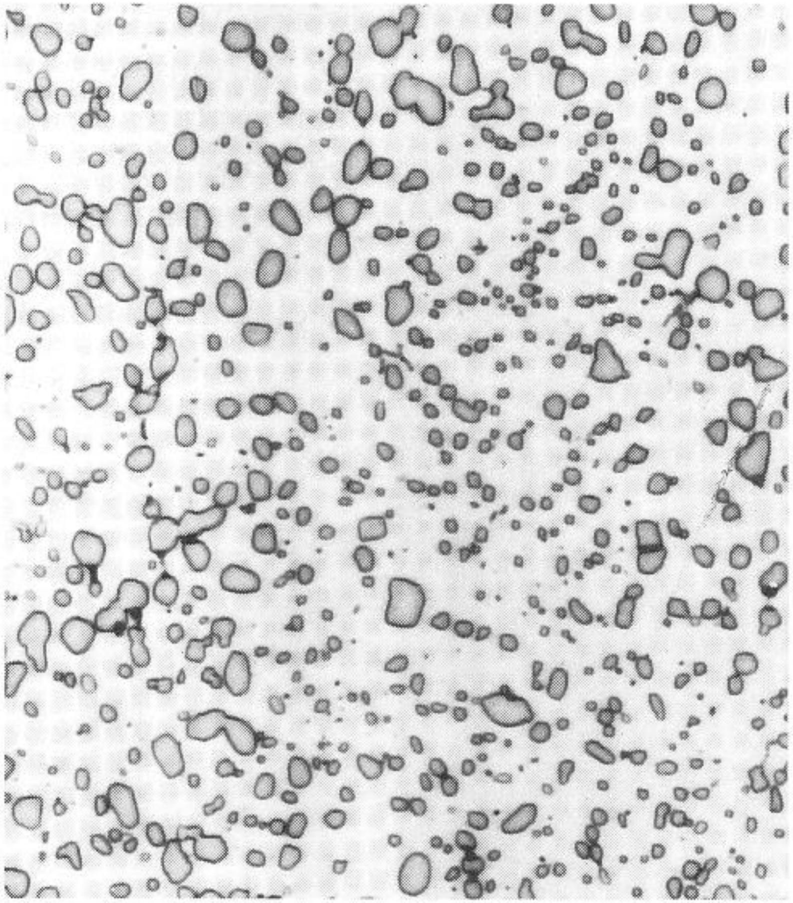
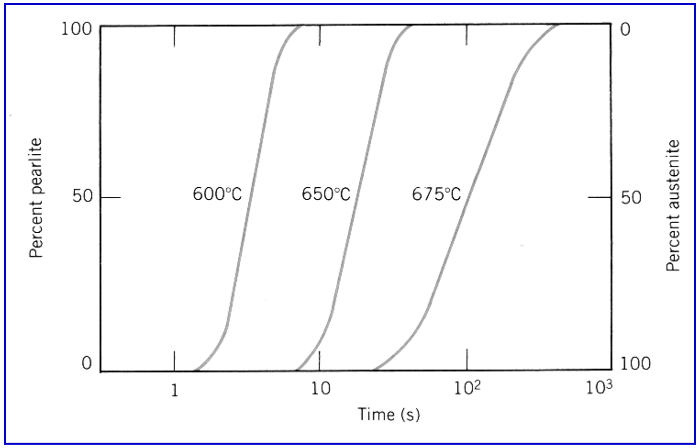


FIGURE 10.10 Photomicrograph of a steel having a spheroidite microstructure. The small particles are cementite; the continuous phase is α ferrite. 1000×.

Isothermal Transformation

FIGURE 10.3 For an iron–carbon alloy of eutectoid composition (0.76 wt% C), isothermal fraction reacted versus the logarithm of time for the austenite-to-pearlite transformation.



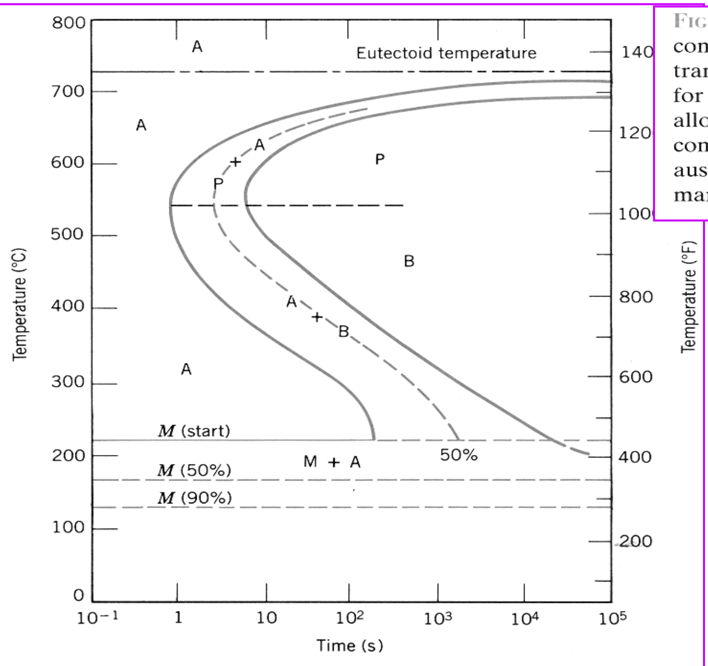


FIGURE 10.13 The complete isothermal transformation diagram for an iron–carbon alloy of eutectoid composition: A, austenite; B, bainite; M, martensite; P, pearlite.

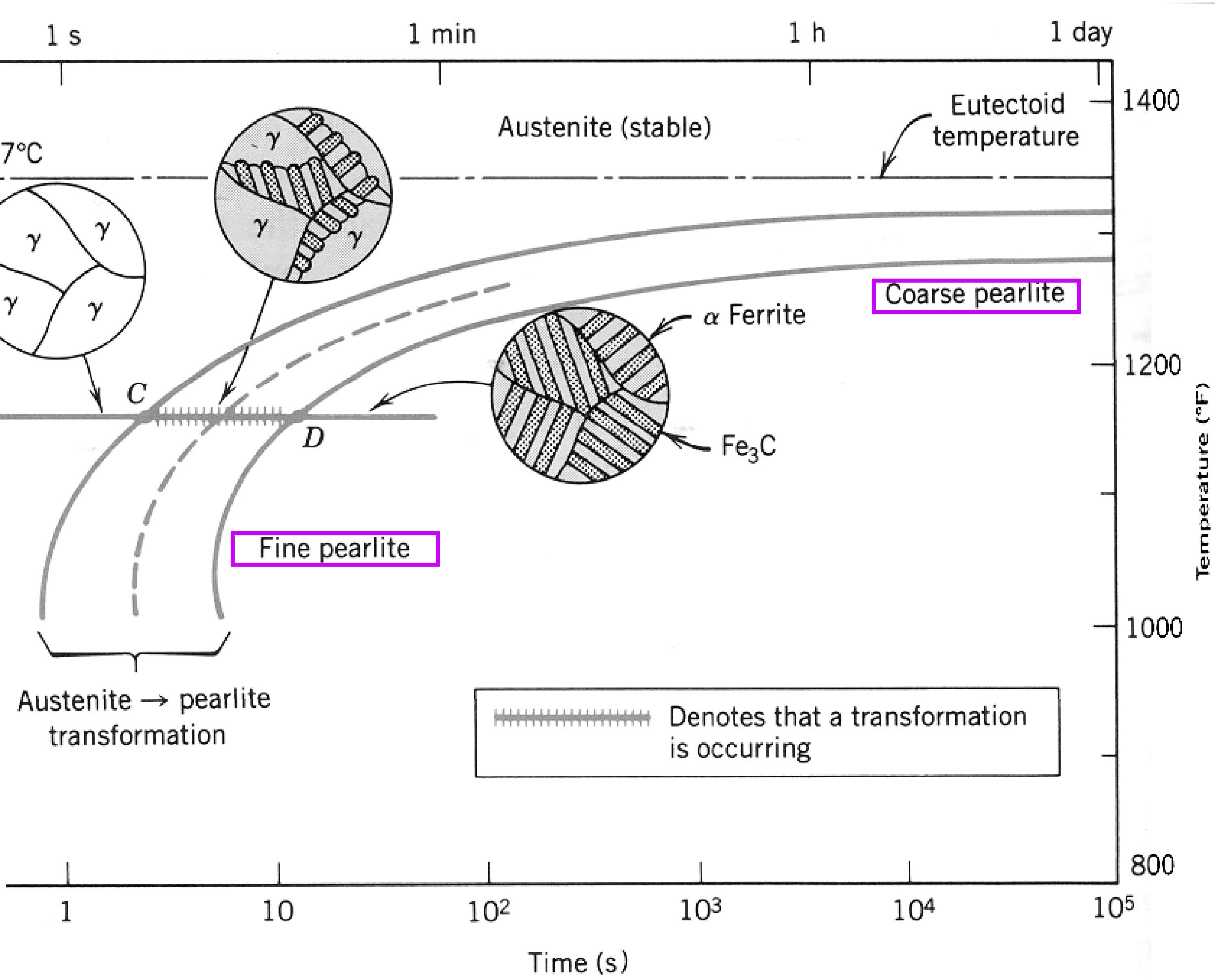
- Conventional heat treatment for producing martensitic steels
- continuous and rapid cooling of an austenitized specimen in some type of quenching medium, such as water, oil, or air

The optimum properties of a steel that has been quenched and then tempered can be realized only if,

- during the quenching heat treatment, the specimen has been converted to a high content of martensite

Isothermal Transformation

FIGURE 10.5 Isothermal transformation diagram for a eutectoid iron–carbon alloy, with superimposed isothermal heat treatment curve (*ABCD*). Microstructures before, during, and after the austenite-to-pearlite transformation are shown.



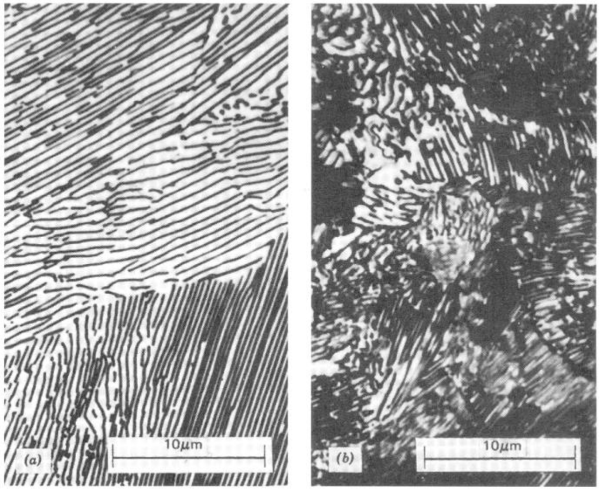


FIGURE 10.6 Photomicrographs of (*a*) coarse pearlite and (*b*) fine pearlite. 3000×.

Bainite and Matensite

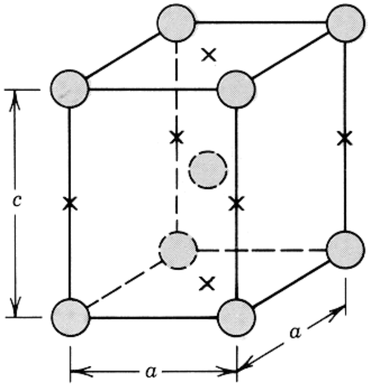


FIGURE 10.11 The body-centered tetragonal unit cell for martensitic steel showing iron atoms (circles) and sites that may be occupied by carbon atoms (crosses). For this tetragonal unit cell, $c > a$.

FIGURE 10.12 Photomicrograph showing the martensitic microstructure. The needle-shaped grains are the martensite phase, and the white regions are austenite that failed to transform during the rapid quench. 1220X.

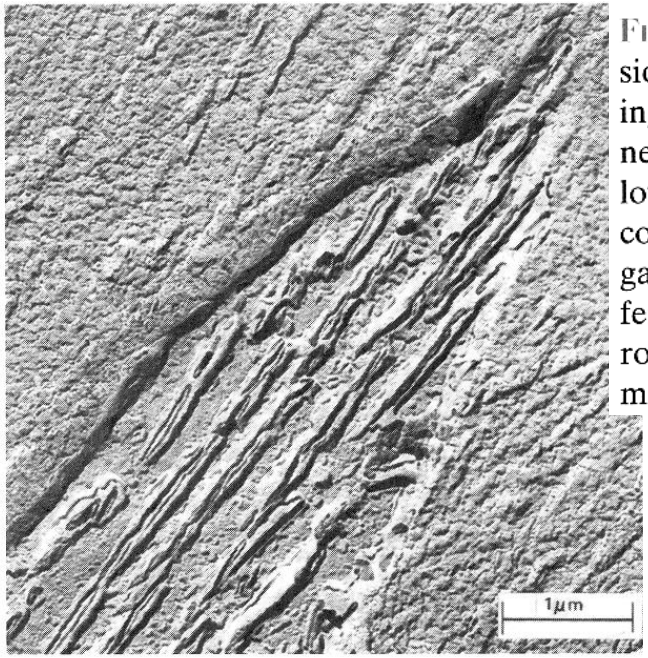


FIGURE 10.8 Replica transmission electron micrograph showing the structure of bainite. A needle of bainite passes from lower left to upper right-hand corners, which consists of elongated particles of Fe_3C within a ferrite matrix. The phase surrounding the bainite needle is martensite.

Strengthening of Steel

- Problem with Martensite M_{50} and M_{90} is that the remaining is austentite
- Which causes instability and cracking
- M100 happens much below room temperature.
- To avoid cracking, alloys are added to steel to increase the M_{100} temperature

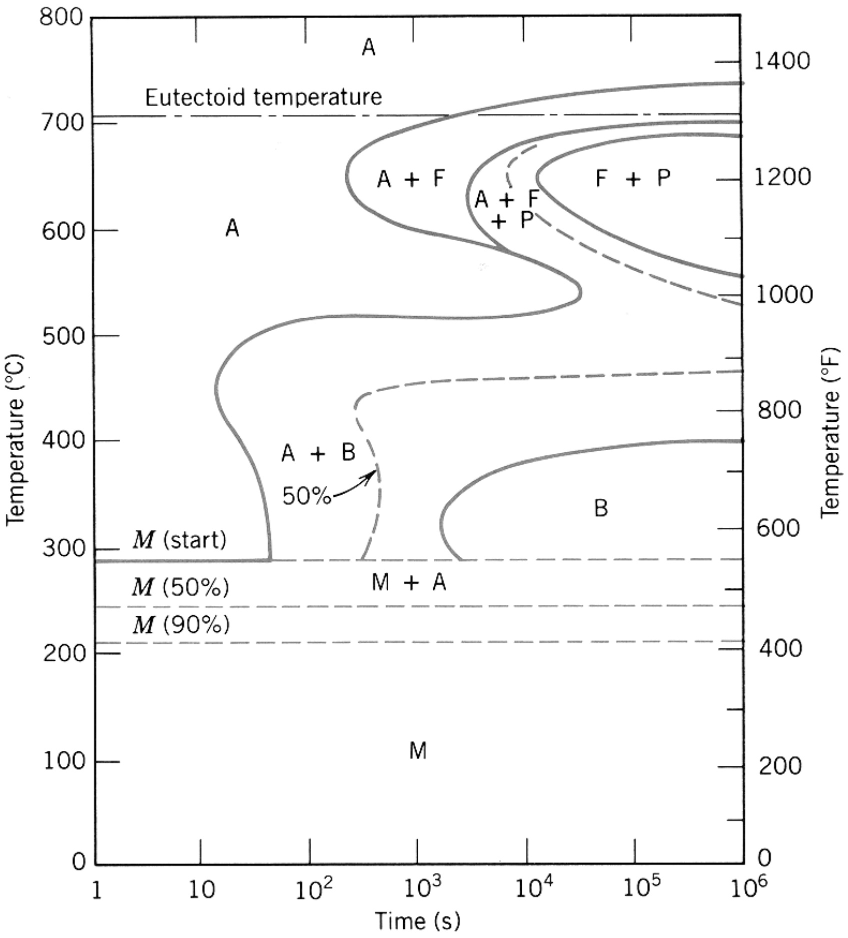


FIGURE 10.14 Isothermal transformation diagram for an alloy steel (type 4340): A, austenite; B, bainite; P, pearlite; M, martensite.

Strengthening of Steel

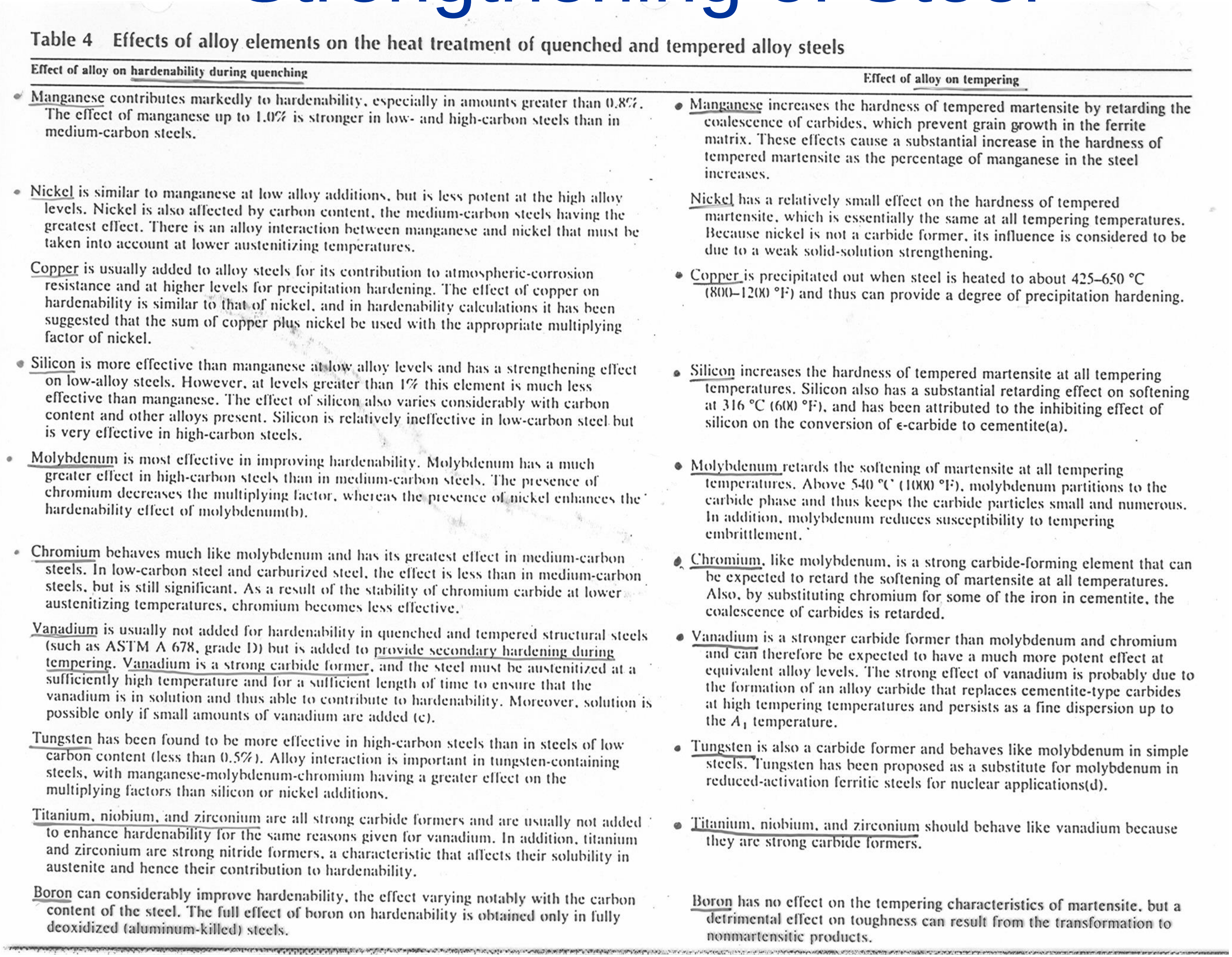
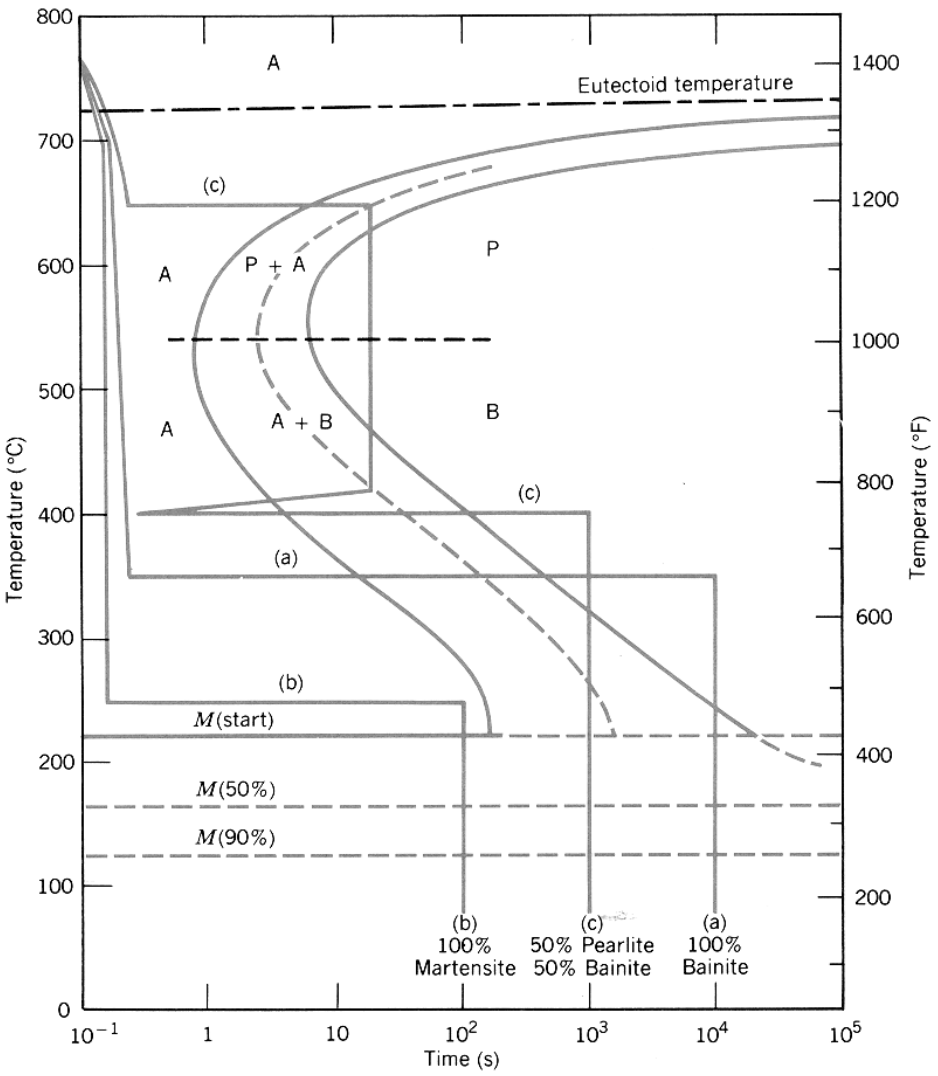


Table 4 Effects of alloy elements on the heat treatment of quenched and tempered alloy steels

Effect of alloy on hardenability during quenching	Effect of alloy on tempering
Manganese contributes markedly to hardenability, especially in amounts greater than 0.8%. The effect of manganese up to 1.0% is stronger in low- and high-carbon steels than in medium-carbon steels.	Manganese increases the hardness of tempered martensite by retarding the coalescence of carbides, which prevent grain growth in the ferrite matrix. These effects cause a substantial increase in the hardness of tempered martensite as the percentage of manganese in the steel increases.
Nickel is similar to manganese at low alloy additions, but is less potent at the high alloy levels. Nickel is also affected by carbon content, the medium-carbon steels having the greatest effect. There is an alloy interaction between manganese and nickel that must be taken into account at lower austenitizing temperatures.	Nickel has a relatively small effect on the hardness of tempered martensite, which is essentially the same at all tempering temperatures. Because nickel is not a carbide former, its influence is considered to be due to a weak solid-solution strengthening.
Copper is usually added to alloy steels for its contribution to atmospheric-corrosion resistance and at higher levels for precipitation hardening. The effect of copper on hardenability is similar to that of nickel, and in hardenability calculations it has been suggested that the sum of copper plus nickel be used with the appropriate multiplying factor of nickel.	Copper is precipitated out when steel is heated to about 425–650 °C (800–1200 °F) and thus can provide a degree of precipitation hardening.
Silicon is more effective than manganese at low alloy levels and has a strengthening effect on low-alloy steels. However, at levels greater than 1% this element is much less effective than manganese. The effect of silicon also varies considerably with carbon content and other alloys present. Silicon is relatively ineffective in low-carbon steel but is very effective in high-carbon steels.	Silicon increases the hardness of tempered martensite at all tempering temperatures. Silicon also has a substantial retarding effect on softening at 316 °C (600 °F), and has been attributed to the inhibiting effect of silicon on the conversion of ϵ-carbide to cementite(a).
Molybdenum is most effective in improving hardenability. Molybdenum has a much greater effect in high-carbon steels than in medium-carbon steels. The presence of chromium decreases the multiplying factor, whereas the presence of nickel enhances the hardenability effect of molybdenum(b).	Molybdenum retards the softening of martensite at all tempering temperatures. Above 540 °C (1000 °F), molybdenum partitions to the carbide phase and thus keeps the carbide particles small and numerous. In addition, molybdenum reduces susceptibility to tempering embrittlement.
Chromium behaves much like molybdenum and has its greatest effect in medium-carbon steels. In low-carbon steel and carburized steel, the effect is less than in medium-carbon steels, but is still significant. As a result of the stability of chromium carbide at lower austenitizing temperatures, chromium becomes less effective.	Chromium, like molybdenum, is a strong carbide-forming element that can be expected to retard the softening of martensite at all temperatures. Also, by substituting chromium for some of the iron in cementite, the coalescence of carbides is retarded.
Vanadium is usually not added for hardenability in quenched and tempered structural steels (such as ASTM A 678, grade D) but is added to provide secondary hardening during tempering. Vanadium is a strong carbide former, and the steel must be austenitized at a sufficiently high temperature and for a sufficient length of time to ensure that the vanadium is in solution and thus able to contribute to hardenability. Moreover, solution is possible only if small amounts of vanadium are added (c).	Vanadium is a stronger carbide former than molybdenum and chromium and can therefore be expected to have a much more potent effect at equivalent alloy levels. The strong effect of vanadium is probably due to the formation of an alloy carbide that replaces cementite-type carbides at high tempering temperatures and persists as a fine dispersion up to the A_1 temperature.
Tungsten has been found to be more effective in high-carbon steels than in steels of low carbon content (less than 0.5%). Alloy interaction is important in tungsten-containing steels, with manganese-molybdenum-chromium having a greater effect on the multiplying factors than silicon or nickel additions.	Tungsten is also a carbide former and behaves like molybdenum in simple steels. Tungsten has been proposed as a substitute for molybdenum in reduced-activation ferritic steels for nuclear applications(d).
Titanium, niobium, and zirconium are all strong carbide formers and are usually not added to enhance hardenability for the same reasons given for vanadium. In addition, titanium and zirconium are strong nitride formers, a characteristic that affects their solubility in austenite and hence their contribution to hardenability.	Titanium, niobium, and zirconium should behave like vanadium because they are strong carbide formers.
Boron can considerably improve hardenability, the effect varying notably with the carbon content of the steel. The full effect of boron on hardenability is obtained only in fully deoxidized (aluminum-killed) steels.	Boron has no effect on the tempering characteristics of martensite, but a detrimental effect on toughness can result from the transformation to nonmartensitic products.

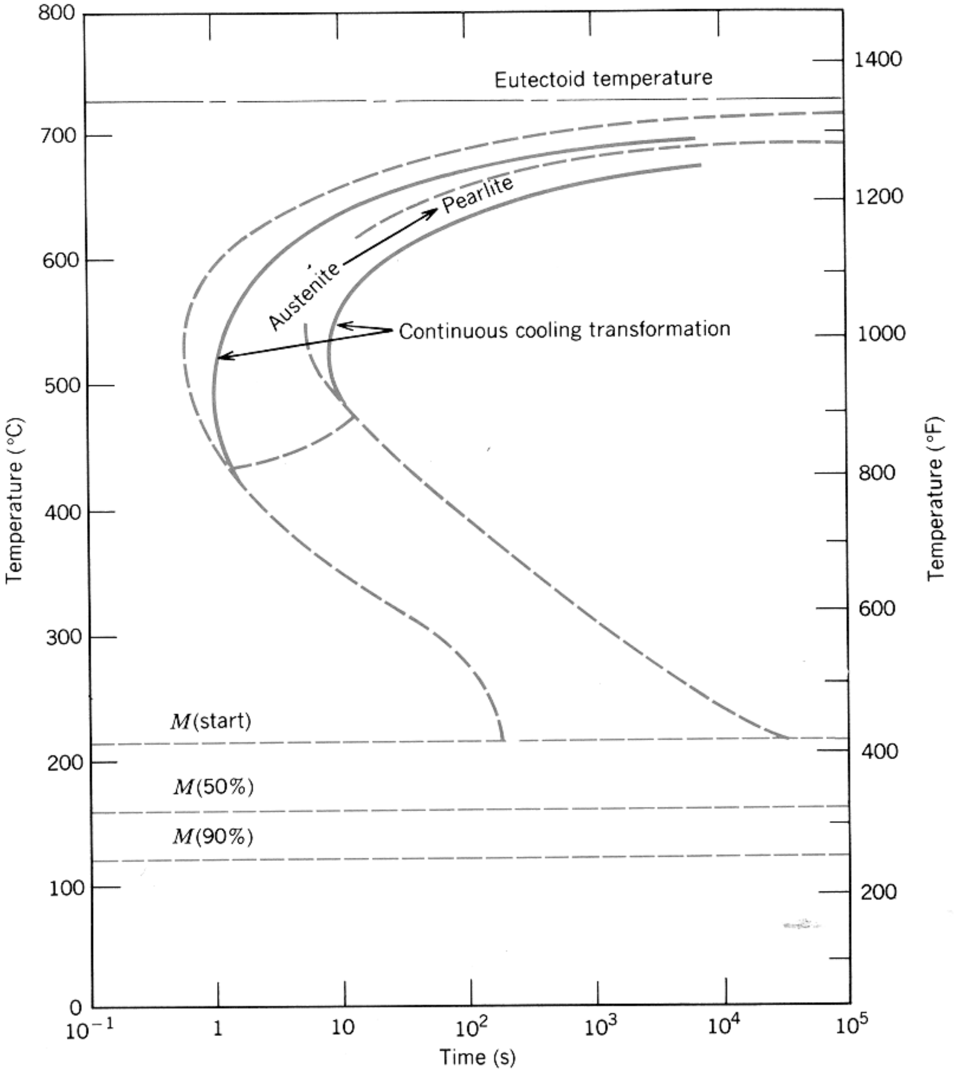
Strengthening of Steel

FIGURE 10.15 Isothermal transformation diagram for an iron–carbon alloy of eutectoid composition and the isothermal heat treatments (a), (b), and (c) in Example Problem 10.1.



Continuous Cooling Curve

FIGURE 10.16 Superimposition of isothermal and continuous cooling transformation diagrams for a eutectoid iron–carbon alloy.



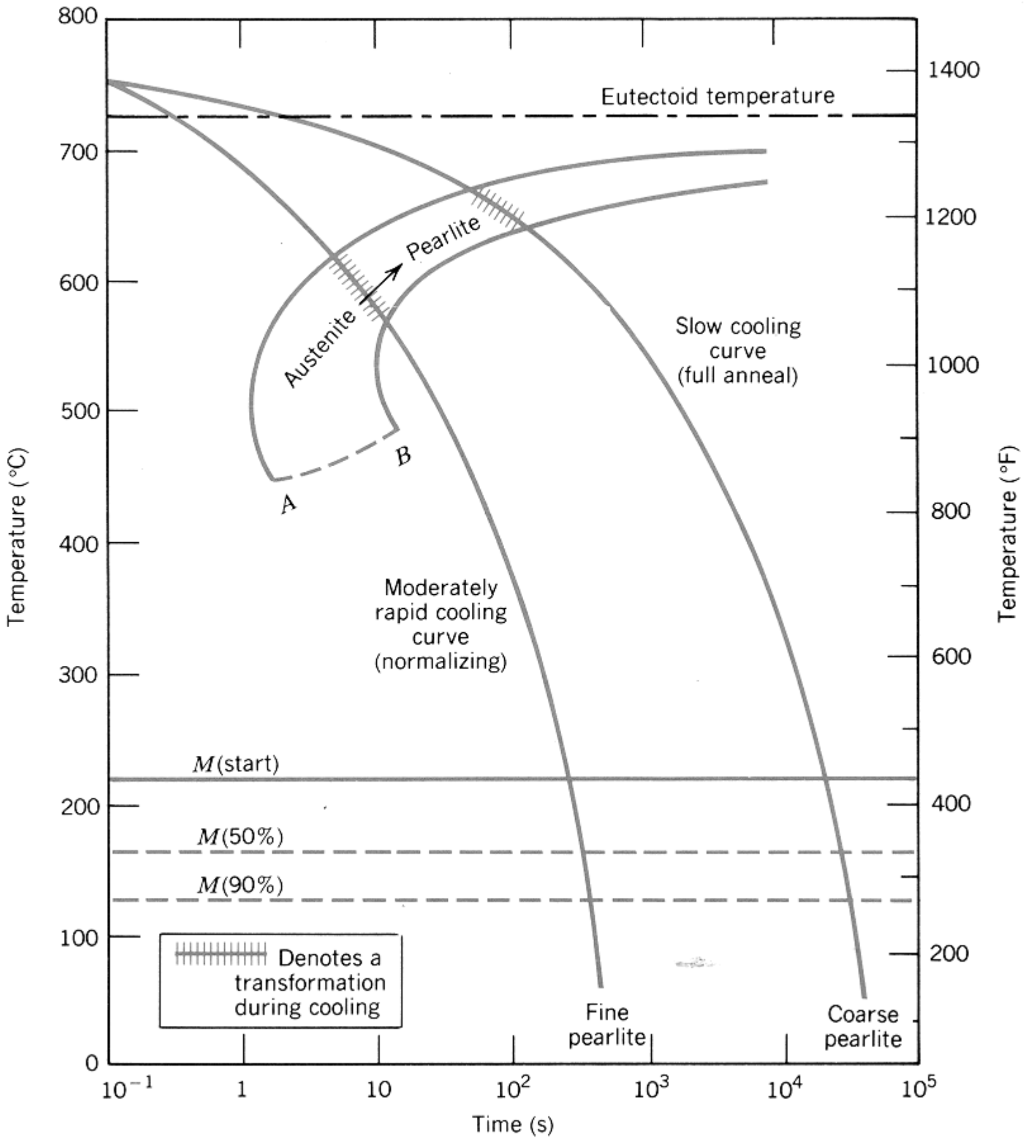
- Isothermal transformation curves are valid only if maintained at same elevate temperature

- It is not practical for heat treatment

- Hence continuous cooling curves are drawn

Continuous Cooling Curve

FIGURE 10.17 Moderately rapid and slow cooling curves superimposed on a continuous cooling transformation diagram for a eutectoid iron–carbon alloy.



Continuous Cooling Curve

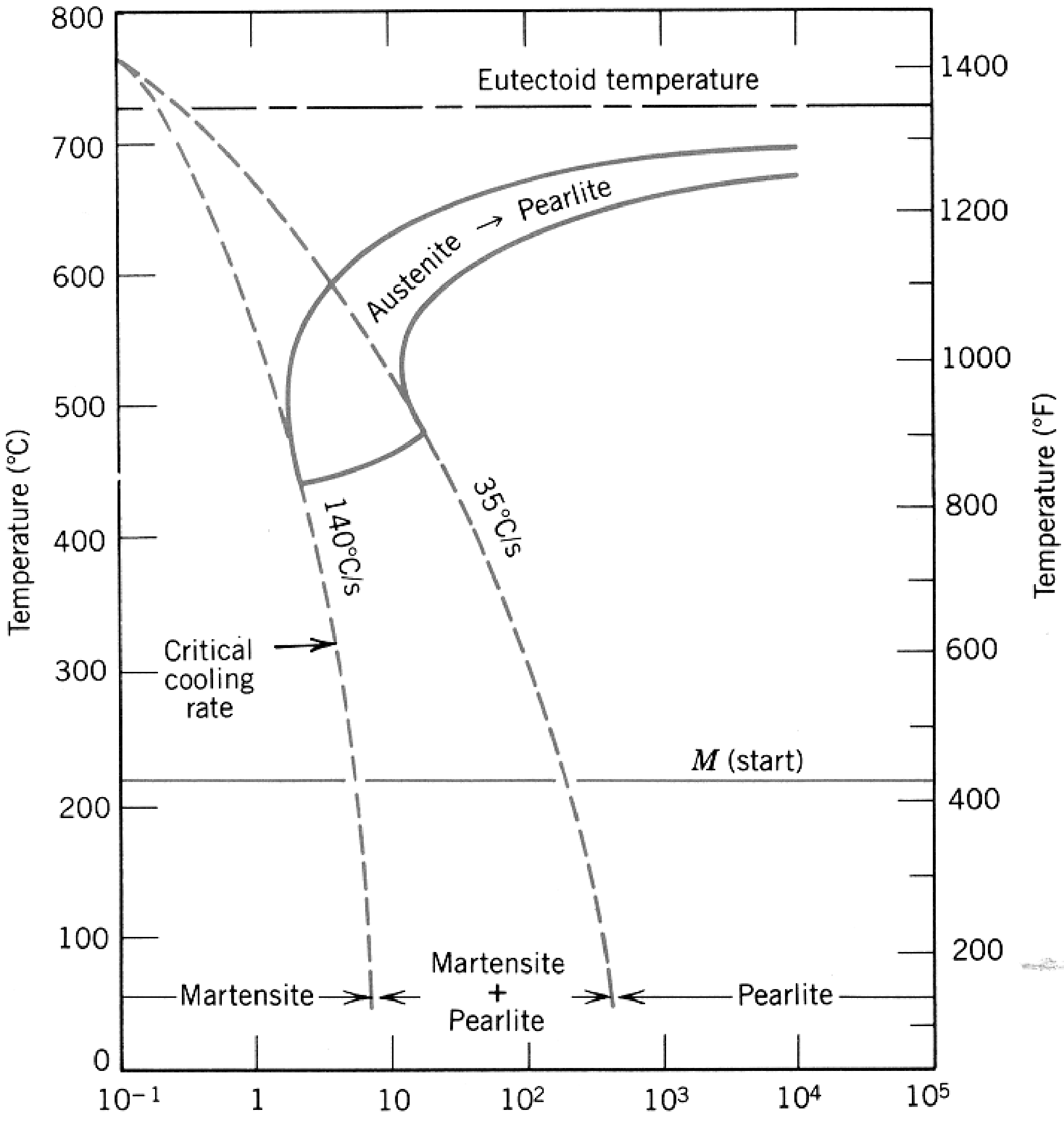


FIGURE 10.18 Continuous cooling transformation diagram for a eutectoid iron–carbon alloy and superimposed cooling curves, demonstrating the dependence of the final mi crostructure on the transformations that oc cur during cooling.

Strength of Pearlite Steel

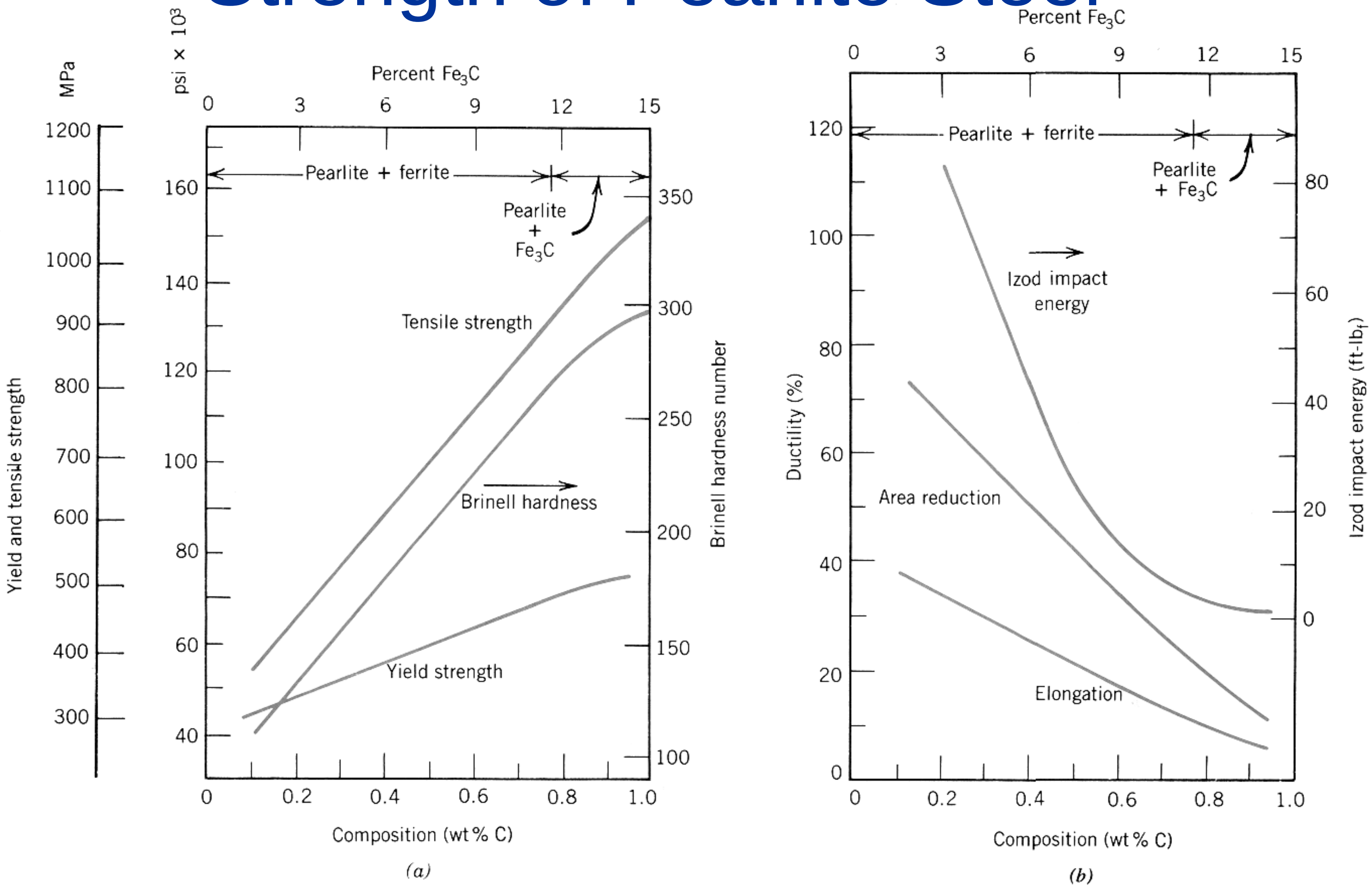
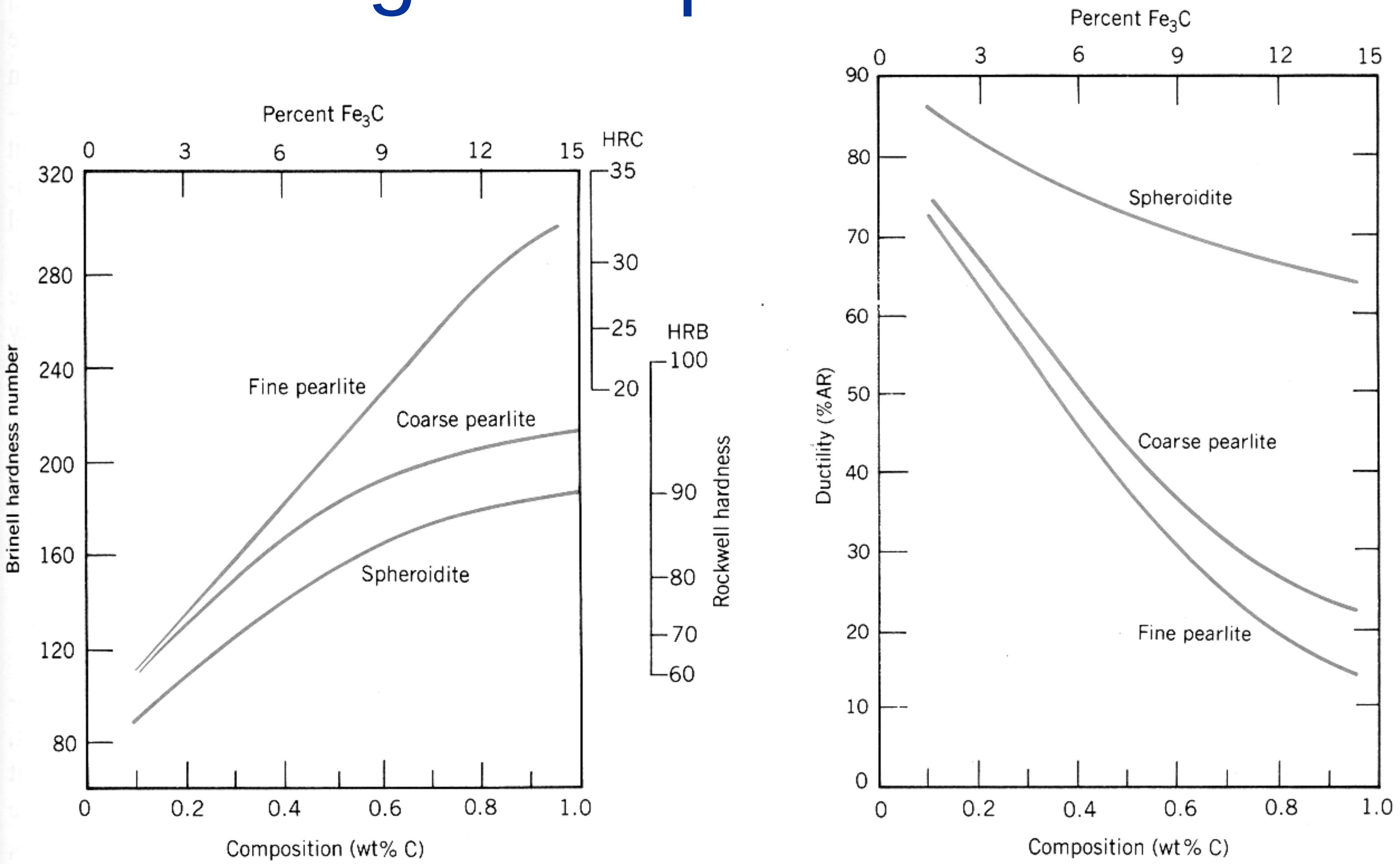


FIGURE 10.20 (*a*) Yield strength, tensile strength, and Brinell hardness versus carbon concentration for plain carbon steels having microstructures consisting of fine pearlite. (*b*) Ductility (%EL and %AR) and Izod impact energy versus carbon concentration for plain carbon steels having microstructures consisting of fine pearlite.

Strength of Spheroidite Steel



(a) *(b)*

FIGURE 10.21 (*a*) Brinell and Rockwell hardness as a function of carbon concentration for plain carbon steels having fine and coarse pearlite as well as spheroidite microstructures. (*b*) Ductility (%AR) as a function of carbon concentration for plain carbon steels having fine and coarse pearlite as well as spheroidite microstructures.

Strength of Martenstic Steel

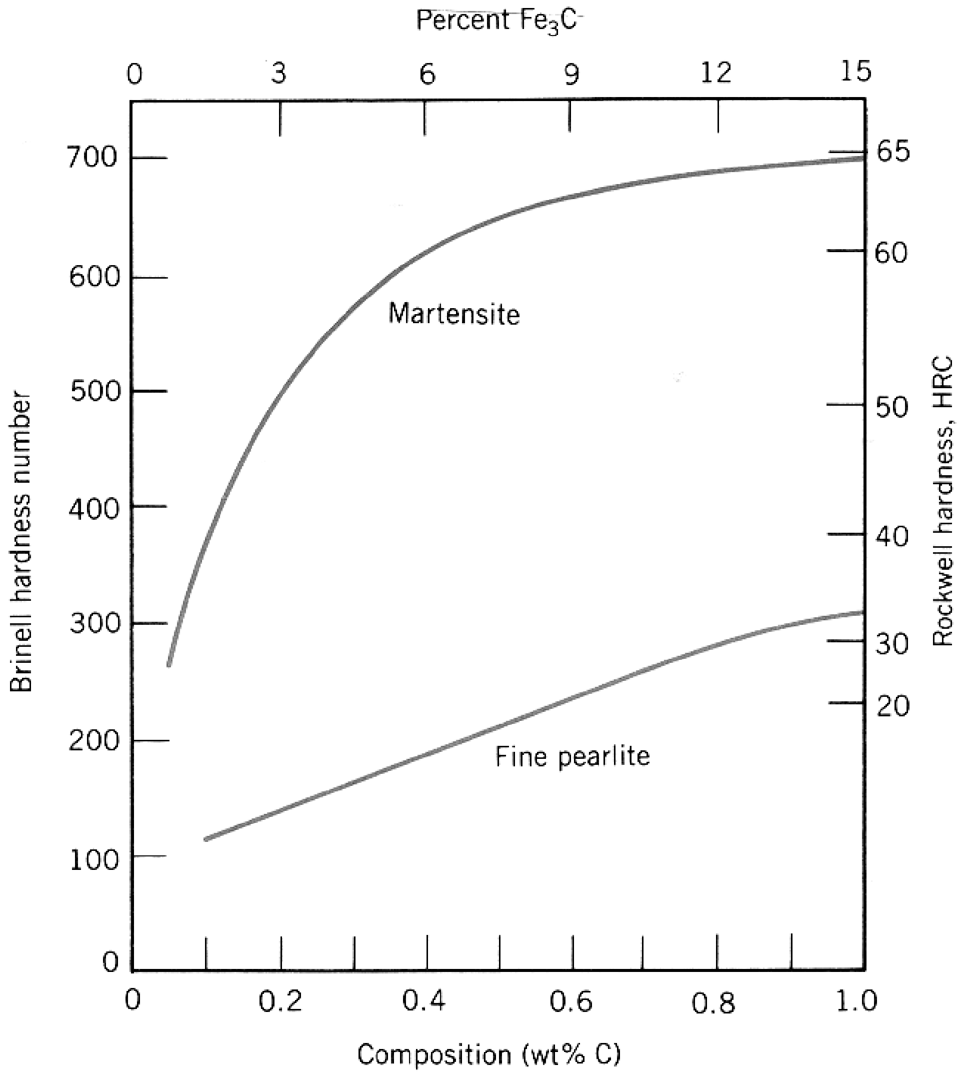


FIGURE 10.22
Hardness as a function of carbon concentration for plain carbon martensitic and fine pearlitic steels.

Quenching & Tempering

- In order to get martensitic steels need continuous, rapid cooling.
- Use quenching medium such as water, oil, air in order to get a high martensite content then temper.
- During cooling, impossible to get uniform cooling rate throughout specimen; surface always cools faster then interior thus variation in microstructure formed.
- Successful heat treating of steels to get predominantly martensite throughout cross section depends mainly on:
  - composition of steel alloy
  - type of quenching medium
  - size and shape of specimen

Tempering Steel

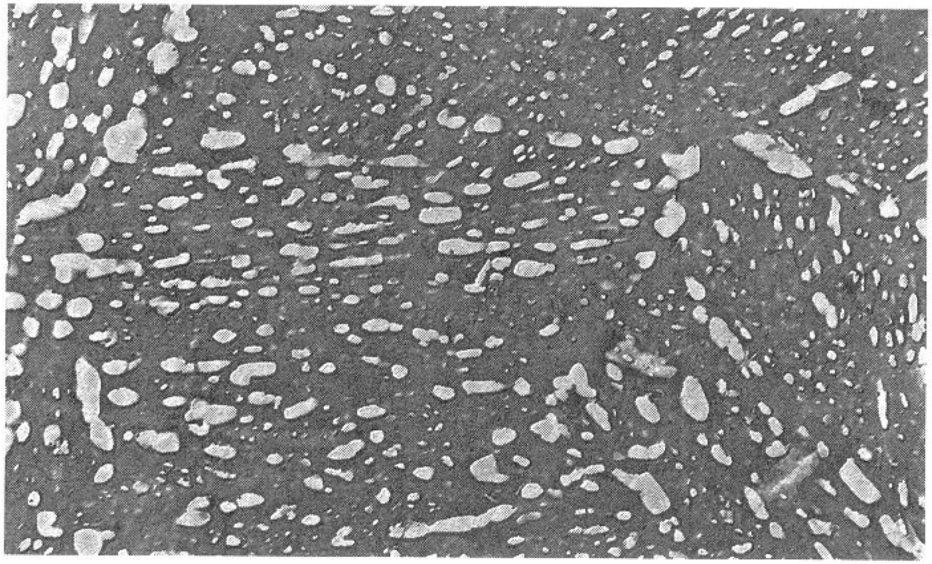
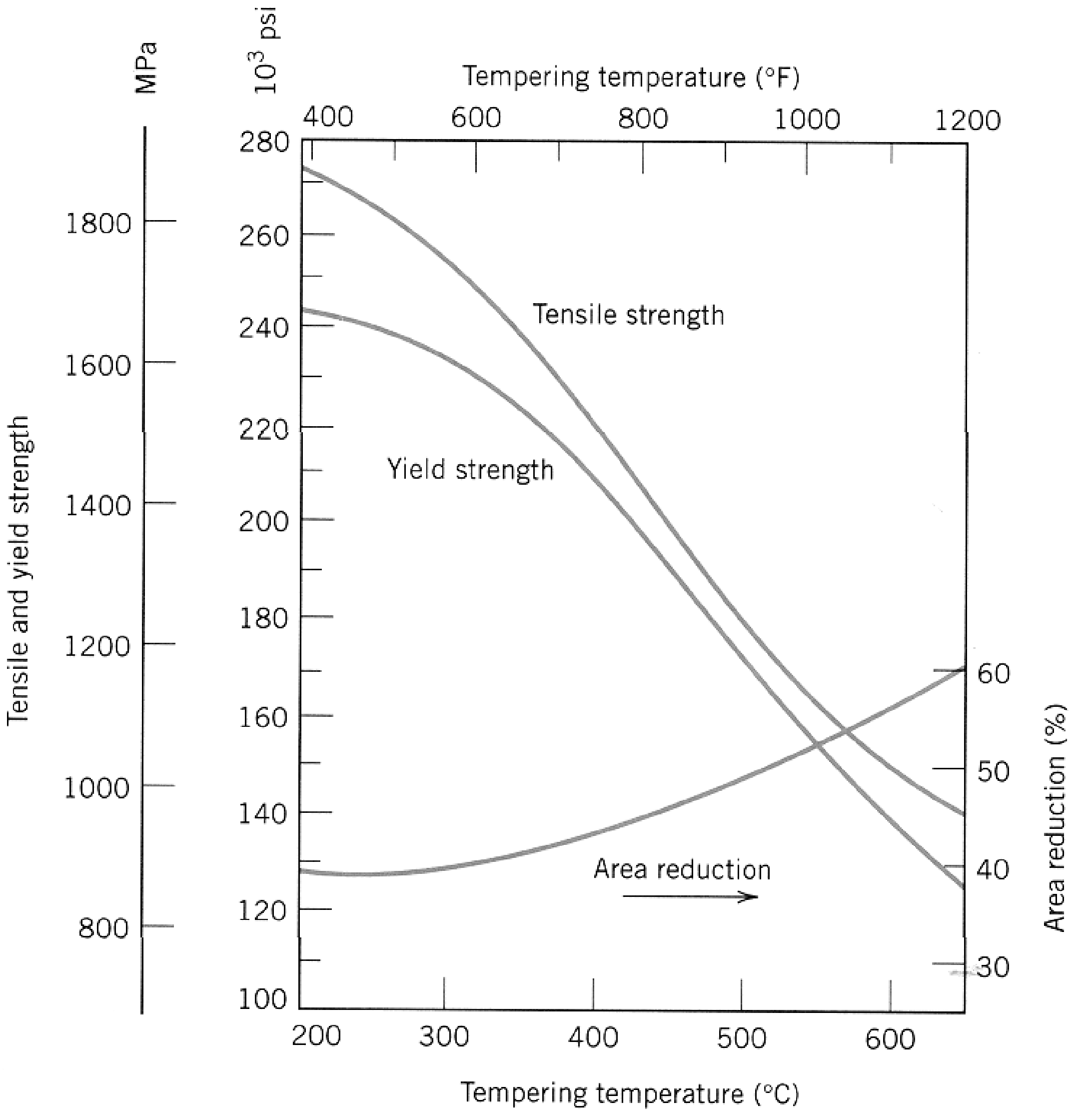


FIGURE 10.23 Electron micrograph of tempered martensite. Tempering was carried out at 594°C (1100°F). The small particles are the cementite phase; the matrix phase is α ferrite. 9300×.

- As quenched Martensite M_{50} has not enough ductility and toughness to be a good engineering structure
- To improve on these martensite is given subsequent heating below eutectoid temperature called tempering
- But strength and hardness decreases with increase in tempering temperature and time

Tempering Steel

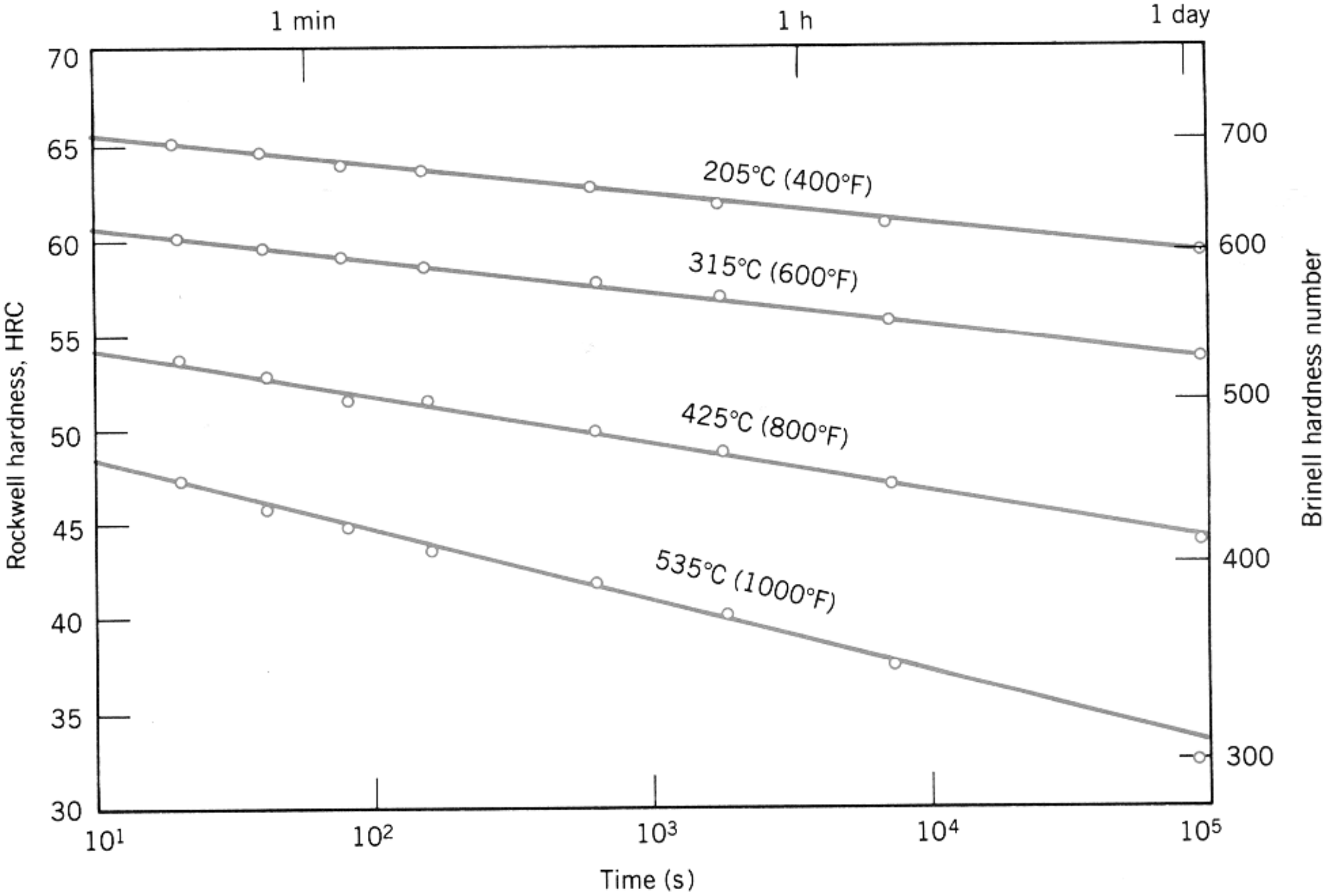
Figure 10.24 Tensile and yield strengths and ductility (%AR) versus tempering temperature for an oil-quenched alloy steel (type 4340).



Tempering Steel

FIGURE 10.25

Hardness versus tempering time for a water-quenched eutectoid plain carbon (1080) steel.



Tempering Alloy Steel

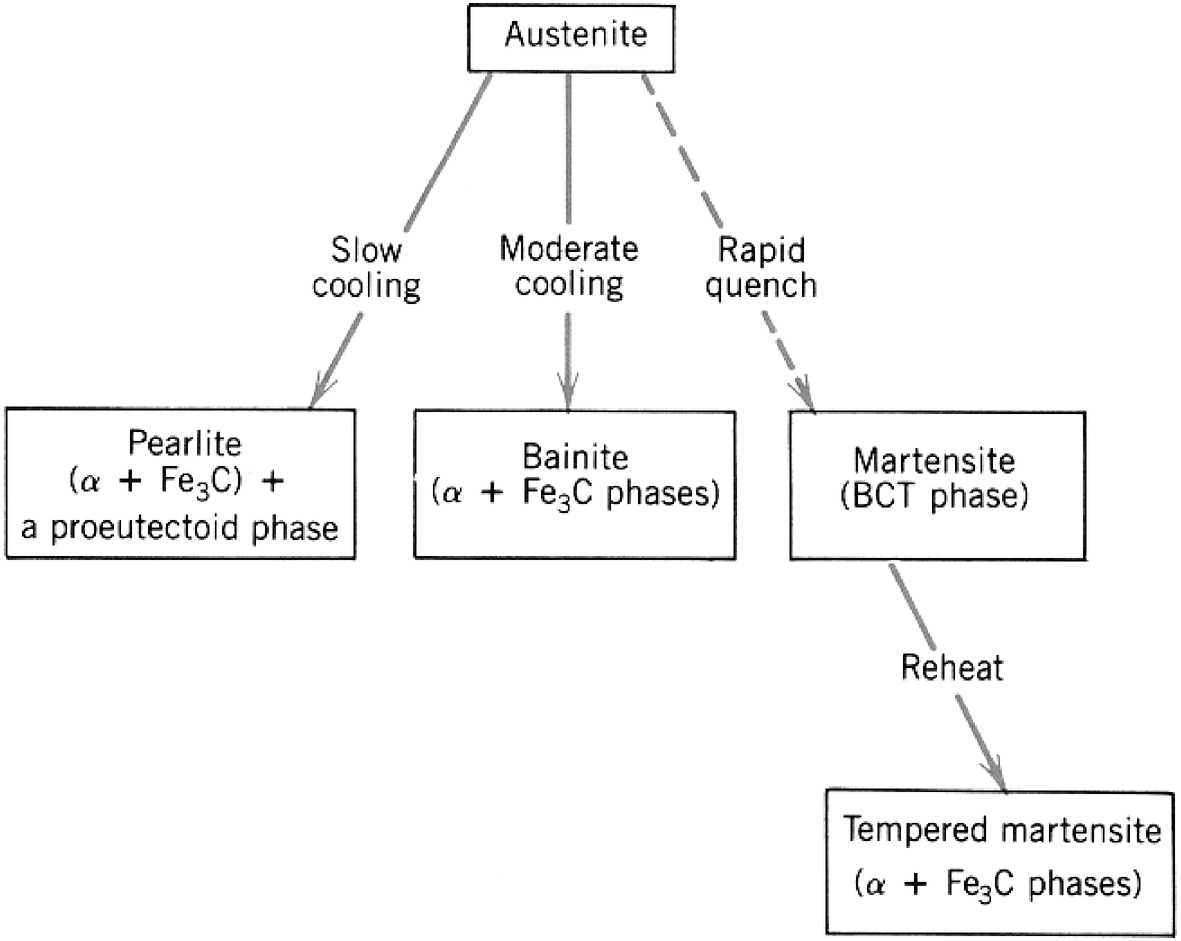


FIGURE 10.26 Possible transformations involving the decomposition of austenite. Solid arrows, transformations involving diffusion; dashed arrow, diffusionless transformation.

Quenching Media

Steel		**Quenchant**
plain carbon steels	-	water
low/med alloyed steels	-	oil
high alloy steels	-	air
Martempering	-	Brine

Severity of quench

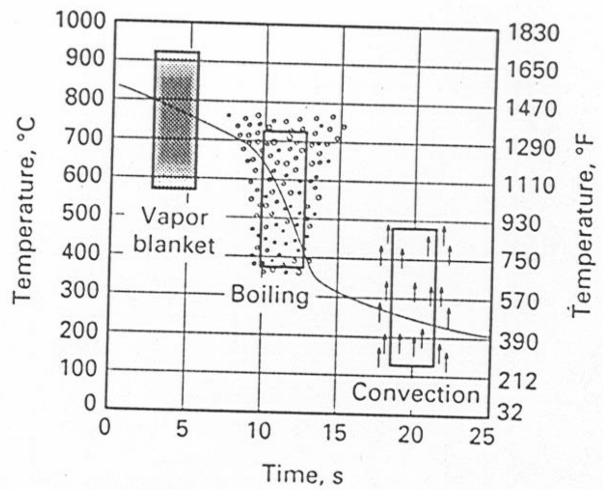


Fig 17 The three stages of quenching. Source: Ref 19

- 3 Stages of Quenching (liquid quenchants)
- Vapour Blanket: cooling medium is vapourized; forms thin "blanket" around sample. Low cooling rate.
- Boiling Stage: vapour no longer sustainable as T dropping; liquid boils on contact to form discrete vapour bubbles that leave surface. Effective heat transfer.
- Convection Stage: Temp is below boiling pt. of liquid, relies on convection of liquid to move heat away – slow Agitation - by pumps/impellors etc.

Defects & Distortions in Heat Treating

- Over-heating & "burning" (low alloy steels)
- Long time at high T causes MnS dissolution & reprecipitation along gbs - intergranular fracture. Occurs during forging/good temp control required.
- Residual Stresses - Heat treatment often causes these.
- - macro: long-range residual stresses, act over large regions compared to grain size, (design of parts).
- - micro: residual (short-range, tenelated stresses), lattice defects, precipitates, about grain size.

Defects & Distortions in Heat Treating

- **Effects of Residual Stresses**
  - dimensional changes, & crack initiation
- dimensional changes often occur when residual stress is eliminated eg. machining.
- Compressive Residual Stresses: Often useful as can reduce effect of imposed tensile stresses (reduce likelihood of fatigue, etc.) These type of residual stresses are often deliberately achieved during processing.
- Tensile Residual Stresses: Undesirable, especially at surface (some heat-treatments especially with phase transformations).
- Control Residual Stresses: By stress-relieving. Grinding of layers.

Residual Stresses Steels

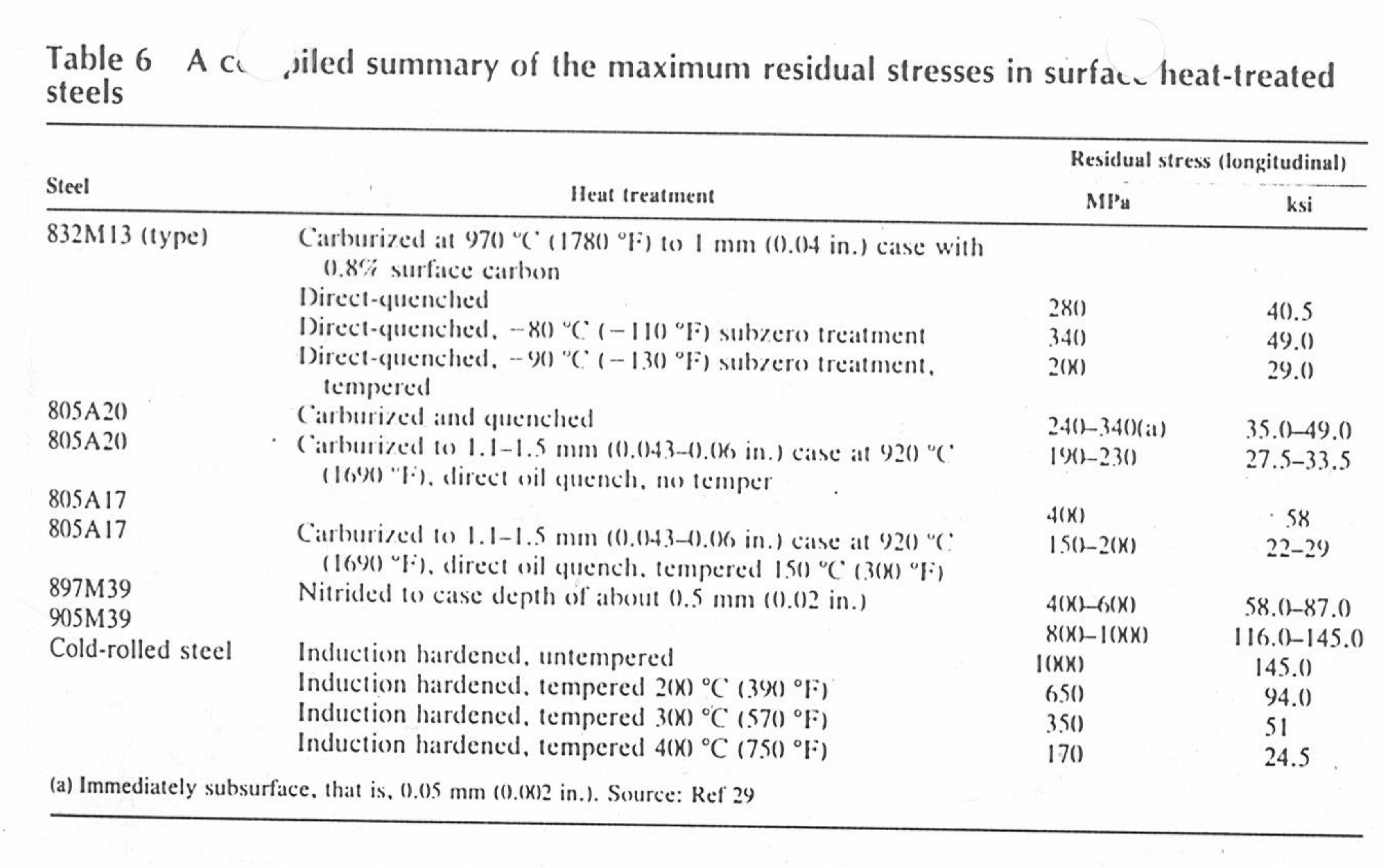


Table 6 A compiled summary of the maximum residual stresses in surface heat-treated steels

Steel	Heat treatment	Residual stress (longitudinal) MPa	Residual stress (longitudinal) ksi
832M13 (type)	Carburized at 970 °C (1780 °F) to 1 mm (0.04 in.) case with 0.8% surface carbon		
	Direct-quenched	280	40.5
	Direct-quenched, −80 °C (−110 °F) subzero treatment	340	49.0
	Direct-quenched, −90 °C (−130 °F) subzero treatment, tempered	200	29.0
805A20	Carburized and quenched	240–340(a)	35.0–49.0
805A20	Carburized to 1.1–1.5 mm (0.043–0.06 in.) case at 920 °C (1690 °F), direct oil quench, no temper	190–230	27.5–33.5
805A17		400	58
805A17	Carburized to 1.1–1.5 mm (0.043–0.06 in.) case at 920 °C (1690 °F), direct oil quench, tempered 150 °C (300 °F)	150–200	22–29
897M39	Nitrided to case depth of about 0.5 mm (0.02 in.)	400–600	58.0–87.0
905M39		800–1000	116.0–145.0
Cold-rolled steel	Induction hardened, untempered	1000	145.0
	Induction hardened, tempered 200 °C (390 °F)	650	94.0
	Induction hardened, tempered 300 °C (570 °F)	350	51
	Induction hardened, tempered 400 °C (750 °F)	170	24.5

(a) Immediately subsurface, that is, 0.05 mm (0.002 in.). Source: Ref 29

Defects & Distortions in Heat Treating

- **Quench Cracking**: - Caused by excessive quenching stresses.

  Due to:

- Part Design: sharp corners, keyways, splines etc. - stress concentrations. Use less severe quench (oil) etc.
- Steel Grade: some grades (higher % c etc) more susceptible
- Part Defects: stringers, inclusions etc.
- Heat-Treating: higher austenitizing temps more likely to cause cracking; coarse grain size; non-uniform cooling, soft spots from inadequate cooling (tongs etc.)

Defects & Distortions in Heat Treating

- **Quench Cracking**: - Caused by excessive quenching stresses.

  Due to:

- Decarburization - changes %C thus changes transformation (CCT) times.

- Warpage: rapid heating/non uniform/quenching residual stresses already present (rolling, grinding etc), uneven hardening & (scale). Long or thin parts.

- Use proper procedures, protect surfaces, fixtures.

Heat Treating Tool Steels

- Usually have high %carbon plus alloying elements for hardness. Cr, V, W, Mo (carbide formers etc.).
- Usually formed first (forged/machined) then heat-treated (not often normalized as air-cooling can cause hardening).
- Quenching medium depends on composition & thickness. Often "hot-quenched" in oil 540º/650ºC
- Tempered (+ often double-tempered to remove untempered martenite from transformation of retained austenite).
- → Quench → M + Retained γ → Temper → MT + M → Temper → MT

Heat Treating Tool Steels

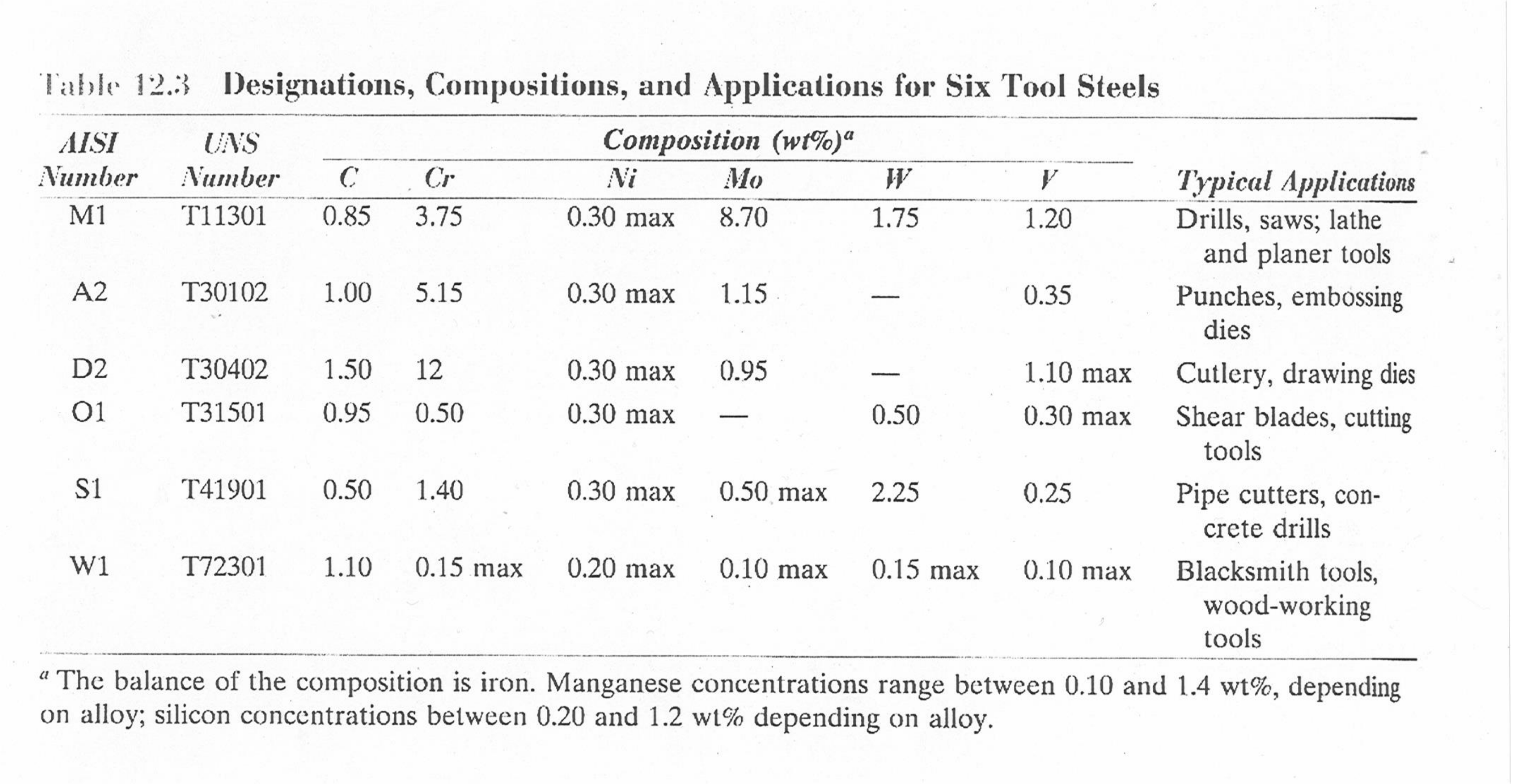
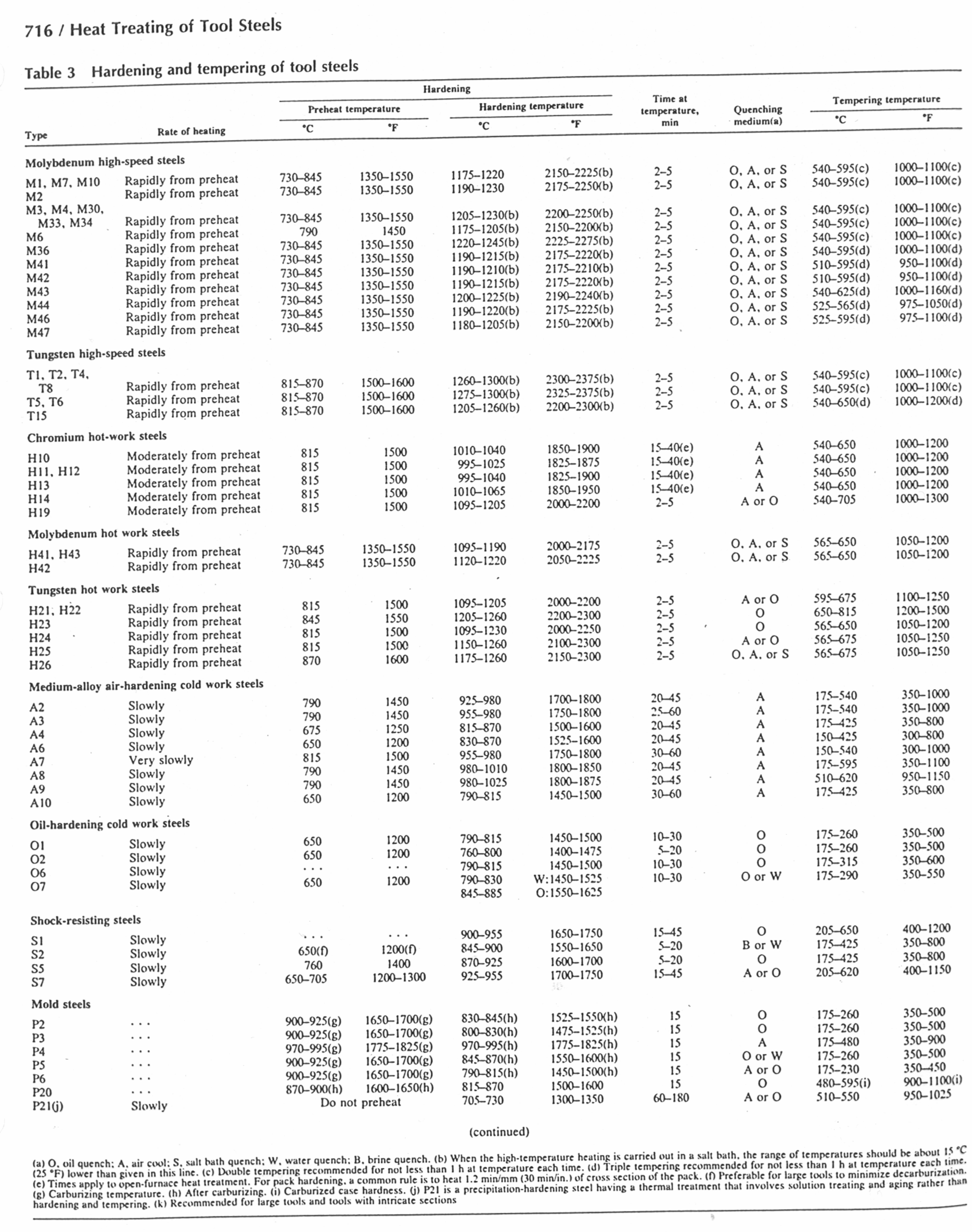


Table 12.3 **Designations, Compositions, and Applications for Six Tool Steels**

AISI Number	*UNS Number*	*Composition (wt%)*[a]						*Typical Applications*
		C	*Cr*	*Ni*	*Mo*	*W*	*V*	
M1	T11301	0.85	3.75	0.30 max	8.70	1.75	1.20	Drills, saws; lathe and planer tools
A2	T30102	1.00	5.15	0.30 max	1.15	—	0.35	Punches, embossing dies
D2	T30402	1.50	12	0.30 max	0.95	—	1.10 max	Cutlery, drawing dies
O1	T31501	0.95	0.50	0.30 max	—	0.50	0.30 max	Shear blades, cutting tools
S1	T41901	0.50	1.40	0.30 max	0.50 max	2.25	0.25	Pipe cutters, concrete drills
W1	T72301	1.10	0.15 max	0.20 max	0.10 max	0.15 max	0.10 max	Blacksmith tools, wood-working tools

[a] The balance of the composition is iron. Manganese concentrations range between 0.10 and 1.4 wt%, depending on alloy; silicon concentrations between 0.20 and 1.2 wt% depending on alloy.

Heat Treating Tool Steels



716 / Heat Treating of Tool Steels

Table 3 Hardening and tempering of tool steels

Type	Rate of heating	Hardening: Preheat temperature °C	Hardening: Preheat temperature °F	Hardening: Hardening temperature °C	Hardening: Hardening temperature °F	Time at temperature, min	Quenching medium(a)	Tempering temperature °C	Tempering temperature °F
Molybdenum high-speed steels									
M1, M7, M10	Rapidly from preheat	730–845	1350–1550	1175–1220	2150–2225(b)	2–5	O, A, or S	540–595(c)	1000–1100(c)
M2	Rapidly from preheat	730–845	1350–1550	1190–1230	2175–2250(b)	2–5	O, A, or S	540–595(c)	1000–1100(c)
M3, M4, M30, M33, M34	Rapidly from preheat	730–845	1350–1550	1205–1230(b)	2200–2250(b)	2–5	O, A, or S	540–595(c)	1000–1100(c)
M6	Rapidly from preheat	790	1450	1175–1205(b)	2150–2200(b)	2–5	O, A, or S	540–595(c)	1000–1100(c)
M36	Rapidly from preheat	730–845	1350–1550	1220–1245(b)	2225–2275(b)	2–5	O, A, or S	540–595(c)	1000–1100(c)
M41	Rapidly from preheat	730–845	1350–1550	1190–1215(b)	2175–2220(b)	2–5	O, A, or S	540–595(d)	1000–1100(d)
M42	Rapidly from preheat	730–845	1350–1550	1190–1210(b)	2175–2210(b)	2–5	O, A, or S	510–595(d)	950–1100(d)
M43	Rapidly from preheat	730–845	1350–1550	1190–1215(b)	2175–2220(b)	2–5	O, A, or S	510–595(d)	950–1100(d)
M44	Rapidly from preheat	730–845	1350–1550	1200–1225(b)	2190–2240(b)	2–5	O, A, or S	540–625(d)	1000–1160(d)
M46	Rapidly from preheat	730–845	1350–1550	1190–1220(b)	2175–2225(b)	2–5	O, A, or S	525–565(d)	975–1050(d)
M47	Rapidly from preheat	730–845	1350–1550	1180–1205(b)	2150–2200(b)	2–5	O, A, or S	525–595(d)	975–1100(d)
Tungsten high-speed steels									
T1, T2, T4, T8	Rapidly from preheat	815–870	1500–1600	1260–1300(b)	2300–2375(b)	2–5	O, A, or S	540–595(c)	1000–1100(c)
T5, T6	Rapidly from preheat	815–870	1500–1600	1275–1300(b)	2325–2375(b)	2–5	O, A, or S	540–595(c)	1000–1100(c)
T15	Rapidly from preheat	815–870	1500–1600	1205–1260(b)	2200–2300(b)	2–5	O, A, or S	540–650(d)	1000–1200(d)
Chromium hot-work steels									
H10	Moderately from preheat	815	1500	1010–1040	1850–1900	15–40(e)	A	540–650	1000–1200
H11, H12	Moderately from preheat	815	1500	995–1025	1825–1875	15–40(e)	A	540–650	1000–1200
H13	Moderately from preheat	815	1500	995–1040	1825–1900	15–40(e)	A	540–650	1000–1200
H14	Moderately from preheat	815	1500	1010–1065	1850–1950	15–40(e)	A	540–650	1000–1200
H19	Moderately from preheat	815	1500	1095–1205	2000–2200	2–5	A or O	540–705	1000–1300
Molybdenum hot work steels									
H41, H43	Rapidly from preheat	730–845	1350–1550	1095–1190	2000–2175	2–5	O, A, or S	565–650	1050–1200
H42	Rapidly from preheat	730–845	1350–1550	1120–1220	2050–2225	2–5	O, A, or S	565–650	1050–1200
Tungsten hot work steels									
H21, H22	Rapidly from preheat	815	1500	1095–1205	2000–2200	2–5	A or O	595–675	1100–1250
H23	Rapidly from preheat	845	1550	1205–1260	2200–2300	2–5	O	650–815	1200–1500
H24	Rapidly from preheat	815	1500	1095–1230	2000–2250	2–5	O	565–650	1050–1200
H25	Rapidly from preheat	815	1500	1150–1260	2100–2300	2–5	A or O	565–675	1050–1250
H26	Rapidly from preheat	870	1600	1175–1260	2150–2300	2–5	O, A, or S	565–675	1050–1250
Medium-alloy air-hardening cold work steels									
A2	Slowly	790	1450	925–980	1700–1800	20–45	A	175–540	350–1000
A3	Slowly	790	1450	955–980	1750–1800	25–60	A	175–540	350–1000
A4	Slowly	675	1250	815–870	1500–1600	20–45	A	175–425	350–800
A6	Slowly	650	1200	830–870	1525–1600	20–45	A	150–425	300–800
A7	Very slowly	815	1500	955–980	1750–1800	30–60	A	150–540	300–1000
A8	Slowly	790	1450	980–1010	1800–1850	20–45	A	175–595	350–1100
A9	Slowly	790	1450	980–1025	1800–1875	20–45	A	510–620	950–1150
A10	Slowly	650	1200	790–815	1450–1500	30–60	A	175–425	350–800
Oil-hardening cold work steels									
O1	Slowly	650	1200	790–815	1450–1500	10–30	O	175–260	350–500
O2	Slowly	650	1200	760–800	1400–1475	5–20	O	175–260	350–500
O6	Slowly	. . .	. . .	790–815	1450–1500	10–30	O	175–315	350–600
O7	Slowly	650	1200	790–830 845–885	W:1450–1525 O:1550–1625	10–30	O or W	175–290	350–550
Shock-resisting steels									
S1	Slowly	. . .	. . .	900–955	1650–1750	15–45	O	205–650	400–1200
S2	Slowly	650(f)	1200(f)	845–900	1550–1650	5–20	B or W	175–425	350–800
S5	Slowly	760	1400	870–925	1600–1700	5–20	O	175–425	350–800
S7	Slowly	650–705	1200–1300	925–955	1700–1750	15–45	A or O	205–620	400–1150
Mold steels									
P2	. . .	900–925(g)	1650–1700(g)	830–845(h)	1525–1550(h)	15	O	175–260	350–500
P3	. . .	900–925(g)	1650–1700(g)	800–830(h)	1475–1525(h)	15	O	175–260	350–500
P4	. . .	970–995(g)	1775–1825(g)	970–995(h)	1775–1825(h)	15	A	175–480	350–900
P5	. . .	900–925(g)	1650–1700(g)	845–870(h)	1550–1600(h)	15	O or W	175–260	350–500
P6	. . .	900–925(g)	1650–1700(g)	790–815(h)	1450–1500(h)	15	A or O	175–230	350–450
P20	. . .	870–900(h)	1600–1650(h)	815–870	1500–1600	15	O	480–595(i)	900–1100(i)
P21(j)	Slowly	Do not preheat		705–730	1300–1350	60–180	A or O	510–550	950–1025

(continued)

(a) O, oil quench; A, air cool; S, salt bath quench; W, water quench; B, brine quench. (b) When the high-temperature heating is carried out in a salt bath, the range of temperatures should be about 15 °C (25 °F) lower than given in this line. (c) Double tempering recommended for not less than 1 h at temperature each time. (d) Triple tempering recommended for not less than 1 h at temperature each time. (e) Times apply to open-furnace heat treatment. For pack hardening, a common rule is to heat 1.2 min/mm (30 min/in.) of cross section of the pack. (f) Preferable for large tools to minimize decarburization. (g) Carburizing temperature. (h) After carburizing. (i) Carburized case hardness. (j) P21 is a precipitation-hardening steel having a thermal treatment that involves solution treating and aging rather than hardening and tempering. (k) Recommended for large tools and tools with intricate sections

Heat Treating Steels

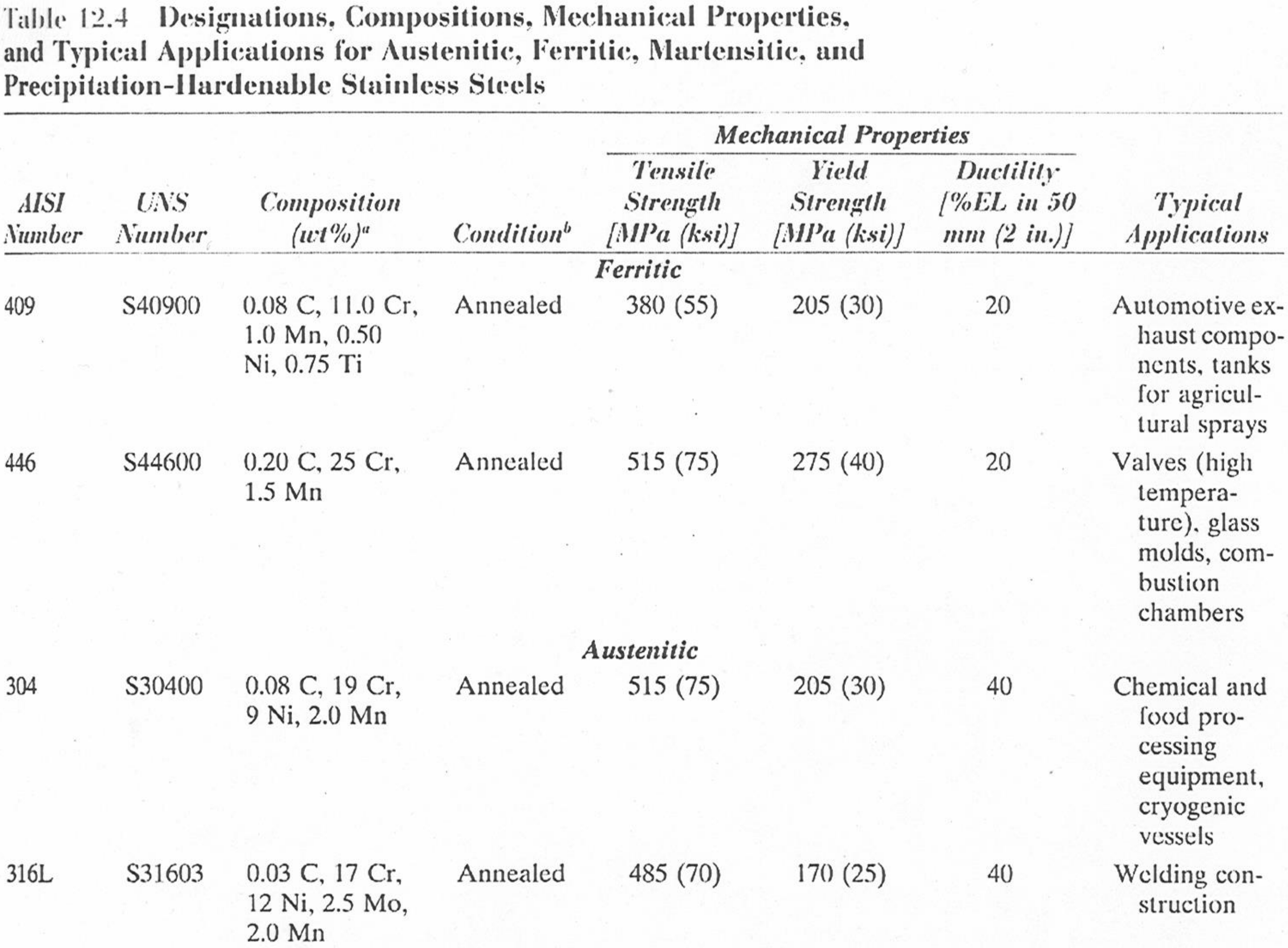


Table 12.4 **Designations, Compositions, Mechanical Properties, and Typical Applications for Austenitic, Ferritic, Martensitic, and Precipitation-Hardenable Stainless Steels**

AISI Number	*UNS Number*	*Composition (wt%)*[a]	*Condition*[b]	*Mechanical Properties* *Tensile Strength [MPa (ksi)]*	*Yield Strength [MPa (ksi)]*	*Ductility [%EL in 50 mm (2 in.)]*	*Typical Applications*
				Ferritic			
409	S40900	0.08 C, 11.0 Cr, 1.0 Mn, 0.50 Ni, 0.75 Ti	Annealed	380 (55)	205 (30)	20	Automotive exhaust components, tanks for agricultural sprays
446	S44600	0.20 C, 25 Cr, 1.5 Mn	Annealed	515 (75)	275 (40)	20	Valves (high temperature), glass molds, combustion chambers
				Austenitic			
304	S30400	0.08 C, 19 Cr, 9 Ni, 2.0 Mn	Annealed	515 (75)	205 (30)	40	Chemical and food processing equipment, cryogenic vessels
316L	S31603	0.03 C, 17 Cr, 12 Ni, 2.5 Mo, 2.0 Mn	Annealed	485 (70)	170 (25)	40	Welding construction

Heat Treating Steels & Alloys

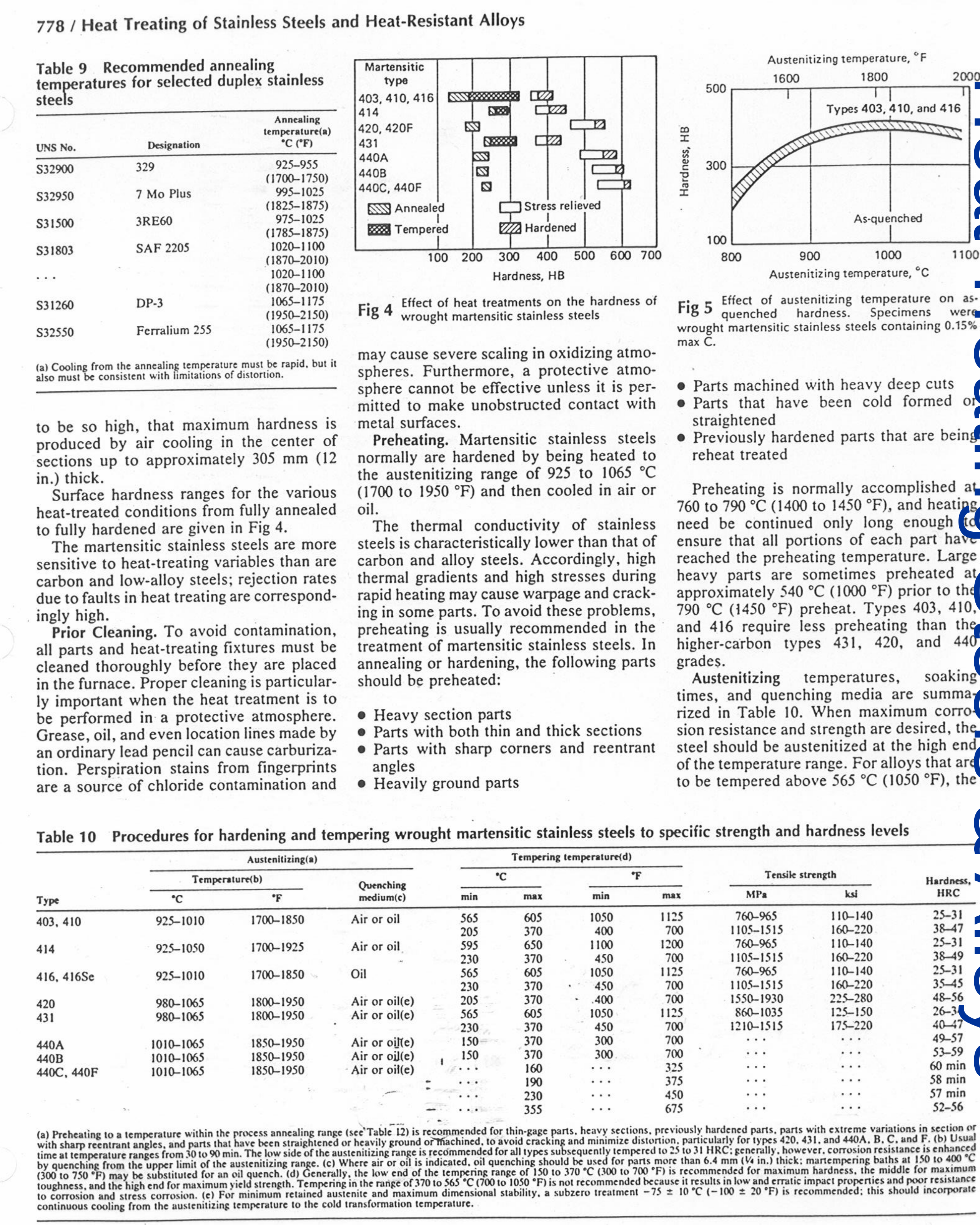


Table 9 Recommended annealing temperatures for selected duplex stainless steels

UNS No.	Designation	Annealing temperature(a) °C (°F)
S32900	329	925–955 (1700–1750)
S32950	7 Mo Plus	995–1025 (1825–1875)
S31500	3RE60	975–1025 (1785–1875)
S31803	SAF 2205	1020–1100 (1870–2010)
. . .		1020–1100 (1870–2010)
S31260	DP-3	1065–1175 (1950–2150)
S32550	Ferralium 255	1065–1175 (1950–2150)

(a) Cooling from the annealing temperature must be rapid, but it also must be consistent with limitations of distortion.

Fig 4 Effect of heat treatments on the hardness of wrought martensitic stainless steels

Fig 5 Effect of austenitizing temperature on as-quenched hardness. Specimens were wrought martensitic stainless steels containing 0.15% max C.

to be so high, that maximum hardness is produced by air cooling in the center of sections up to approximately 305 mm (12 in.) thick.

Surface hardness ranges for the various heat-treated conditions from fully annealed to fully hardened are given in Fig 4.

The martensitic stainless steels are more sensitive to heat-treating variables than are carbon and low-alloy steels; rejection rates due to faults in heat treating are correspondingly high.

Prior Cleaning. To avoid contamination, all parts and heat-treating fixtures must be cleaned thoroughly before they are placed in the furnace. Proper cleaning is particularly important when the heat treatment is to be performed in a protective atmosphere. Grease, oil, and even location lines made by an ordinary lead pencil can cause carburization. Perspiration stains from fingerprints are a source of chloride contamination and may cause severe scaling in oxidizing atmospheres. Furthermore, a protective atmosphere cannot be effective unless it is permitted to make unobstructed contact with metal surfaces.

Preheating. Martensitic stainless steels normally are hardened by being heated to the austenitizing range of 925 to 1065 °C (1700 to 1950 °F) and then cooled in air or oil.

The thermal conductivity of stainless steels is characteristically lower than that of carbon and alloy steels. Accordingly, high thermal gradients and high stresses during rapid heating may cause warpage and cracking in some parts. To avoid these problems, preheating is usually recommended in the treatment of martensitic stainless steels. In annealing or hardening, the following parts should be preheated:

- Heavy section parts
- Parts with both thin and thick sections
- Parts with sharp corners and reentrant angles
- Heavily ground parts
- Parts machined with heavy deep cuts
- Parts that have been cold formed or straightened
- Previously hardened parts that are being reheat treated

Preheating is normally accomplished at 760 to 790 °C (1400 to 1450 °F), and heating need be continued only long enough to ensure that all portions of each part have reached the preheating temperature. Large heavy parts are sometimes preheated at approximately 540 °C (1000 °F) prior to the 790 °C (1450 °F) preheat. Types 403, 410, and 416 require less preheating than the higher-carbon types 431, 420, and 440 grades.

Austenitizing temperatures, soaking times, and quenching media are summarized in Table 10. When maximum corrosion resistance and strength are desired, the steel should be austenitized at the high end of the temperature range. For alloys that are to be tempered above 565 °C (1050 °F), the

Table 10 Procedures for hardening and tempering wrought martensitic stainless steels to specific strength and hardness levels

Type	Austenitizing(a)			Tempering temperature(d)				Tensile strength		Hardness, HRC
	Temperature(b)		Quenching medium(c)	°C		°F				
	°C	°F		min	max	min	max	MPa	ksi	
403, 410	925–1010	1700–1850	Air or oil	565	605	1050	1125	760–965	110–140	25–31
				205	370	400	700	1105–1515	160–220	38–47
414	925–1050	1700–1925	Air or oil	595	650	1100	1200	760–965	110–140	25–31
				230	370	450	700	1105–1515	160–220	38–49
416, 416Se	925–1010	1700–1850	Oil	565	605	1050	1125	760–965	110–140	25–31
				230	370	450	700	1105–1515	160–220	35–45
420	980–1065	1800–1950	Air or oil(e)	205	370	400	700	1550–1930	225–280	48–56
431	980–1065	1800–1950	Air or oil(e)	565	605	1050	1125	860–1035	125–150	26–3
				230	370	450	700	1210–1515	175–220	40–47
440A	1010–1065	1850–1950	Air or oil(e)	150	370	300	700	. . .	. . .	49–57
440B	1010–1065	1850–1950	Air or oil(e)	150	370	300	700	. . .	. . .	53–59
440C, 440F	1010–1065	1850–1950	Air or oil(e)	. . .	160	. . .	325	. . .	. . .	60 min
				. . .	190	. . .	375	. . .	. . .	58 min
				. . .	230	. . .	450	. . .	. . .	57 min
				. . .	355	. . .	675	. . .	. . .	52–56

(a) Preheating to a temperature within the process annealing range (see Table 12) is recommended for thin-gage parts, heavy sections, previously hardened parts, parts with extreme variations in section or with sharp reentrant angles, and parts that have been straightened or heavily ground or machined, to avoid cracking and minimize distortion, particularly for types 420, 431, and 440A, B, C, and F. (b) Usual time at temperature ranges from 30 to 90 min. The low side of the austenitizing range is recommended for all types subsequently tempered to 25 to 31 HRC; generally, however, corrosion resistance is enhanced by quenching from the upper limit of the austenitizing range. (c) Where air or oil is indicated, oil quenching should be used for parts more than 6.4 mm (¼ in.) thick; martempering baths at 150 to 400 °C (300 to 750 °F) may be substituted for an oil quench. (d) Generally, the low end of the tempering range of 150 to 370 °C (300 to 700 °F) is recommended for maximum hardness, the middle for maximum toughness, and the high end for maximum yield strength. Tempering in the range of 370 to 565 °C (700 to 1050 °F) is not recommended because it results in low and erratic impact properties and poor resistance to corrosion and stress corrosion. (e) For minimum retained austenite and maximum dimensional stability, a subzero treatment −75 ± 10 °C (−100 ± 20 °F) is recommended; this should incorporate continuous cooling from the austenitizing temperature to the cold transformation temperature.

Temper Embrittlement

- The toughness of some steels can be reduced by tempering at certain temperatures (between 375 and 575°C and slow cooling). Usually due to presence of impurities (Mn, Ni, Cr, Sb, P, As, Sn).
- Avoid temper embrittlement by:
- 1) controlling composition
- 2) Temper above 575°C or below 375°C followed by fast cooling