High-Strength, Low-Alloy Steels

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Summary. High-strength, low-alloy (HSLA) steels have nearly the same composition as plain carbon steels. However, they are up to twice as strong and their greater load-bearing capacity allows engineering use in lighter sections. Their high strength is derived from a combination of grain refinement; precipitation strengthening due to minor additions of vanadium, niobium, or titanium; and modifications of manufacturing processes, such as controlled rolling and controlled cooling of otherwise essentially plain carbon steel. HSLA steels are less formable than lower strength steels, but dual-phase steels, which evolved from HSLA steels, have ferrite-martensite microstructures and better formability than HSLA steels of similar strength. This improved formability has substantially increased the utilization potential of high-strength steels in the manufacture of complex components. This article reviews the development of HSLA and dual-phase steels and discusses the effects of variations in microstructure and chemistry on their mechanical properties.

High-strength, low-alloy (HSLA) steels represent a new class of materials which evolved from plain carbon or mild steel. They are the precursors of dual-phase steels. HSLA steels have a balance of mechanical properties and provide suitable replacements for other steels in many applications. Their development was triggered by economic considerations and societal need, but is based on solid technical foundations (*1*).

At the turn of the century, when the first horseless carriages were appearing on the scene, plain carbon sheet steel was not a significant product of the steel industry. At that time steel was being used mainly for its strength, rigidity, and ductility in applications such as railroad equipment, machinery, and weapons. But the growth of consumer goods industries, especially automobiles, changed this. These industries found that sheet steel not only provided strength and rigidity but also could be formed easily into the intricate shapes desired for consumer products. Also, sheet steel had an attractive surface after forming, was easily joined to other parts, and was available at reasonable cost.

Metallurgically speaking, plain carbon sheet steels include all flat-rolled products less than 5.84 millimeters thick that contain less than 0.15 percent carbon and less than 1 percent other elements such as manganese and silicon, with the balance being iron. Their typical mechanical properties are yield and tensile strength in the ranges 150 to 200 and 280 to 350 megapascals, respectively, and total elongation or ductility of 30 to 40 percent in a 50.8-mm gage length. Yield strength (YS) is defined as the load needed per unit cross-sectional area of the steel to produce plastic flow and is usually measured at 0.2 percent strain. Ultimate tensile strength (UTS) is the maximum load-carrying ability of the steel per unit cross-sectional area. The strain at UTS is referred to as the uniform elongation (e_u); the strain to failure is called total elongation (e_T). Both e_u and e_T are measures of steel ductility or formability, the two latter terms being used synonymously.

The microstructure of plain carbon steel consists of large ferrite crystals or grains, with grain boundary iron carbides and a small volume fraction of pearlite (Fig. 1). Ferrite is almost pure iron, is also known as α-iron, and has a low solubility for carbon. The grain boundary carbide composition is iron carbide (Fe_3C), which is also known as cementite. Pearlite is characterized by a lamellar structure consisting of alternate layers of ferrite and cementite and forms in the steel under certain cooling conditions.

As sheet steel usage grew, stronger steels with different properties were needed for new applications or to improve durability or manufacturability in existing applications. In the early 1900's high strength was developed in hot-rolled steels by alloying with carbon, sometimes as much as 0.3 percent. However, mechanical properties such as fracture toughness, weldability, and formability suffer with increasing carbon content, and other ways of strengthening low-carbon steel had to be developed. Several research breakthroughs in the 1950's and 1960's triggered the development of present-day HSLA steels (*2*). These advances were a result of an improved understanding of the effect of interactions between alloying elements, such as vanadium, niobium, and titanium, and processing variables on grain refinement, steel strength, and ductility. Vanadium, niobium, and titanium are also referred to as microalloying elements in HSLA steels because of the small quantity of addition required. Methods for controlling the shape of nonmetallic sulfide inclusions in the steel also evolved. All steels contain nonmetallic inclusions such as oxides and sulfides as a consequence of the steelmaking process. Steel mechanical properties can be improved by controlling the shape of these inclusions.

Like plain carbon steel, HSLA steels contain less than 0.15 percent carbon. In addition, they typically contain about 1.0 percent manganese, about 0.1 percent microalloying elements, less than 0.6 percent silicon, and about 0.1 percent alloying elements for inclusion shape control, the balance being iron. The microstructural constituents of HSLA steels are similar to those of plain carbon steel, but the ferrite grain size is far smaller (Fig. 2) because of the microalloying addition and special processing. High strength is developed in these steels by grain refinement, substitutional strengthening due to manganese and silicon, and precipitation strengthening by microalloy precipitates.

HSLA steels are currently produced with the following typical mechanical properties: yield strengths, 350 to 700 MPa; ultimate tensile strength, 450 to 850 MPa; and total elongation, 14 to 27 percent. They are high-strength only in comparison with plain carbon steel, since quenched and tempered high-carbon steels may have tensile strengths as high as 1100 MPa. The reason for this limitation in strength is that other properties, such as fracture toughness, formability, weldability, and fatigue resistance, and processing considerations must also be weighed in the balance. Of-

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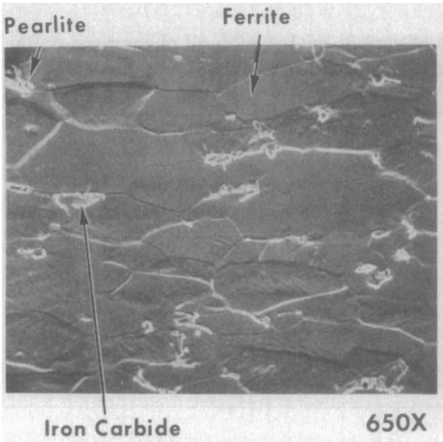


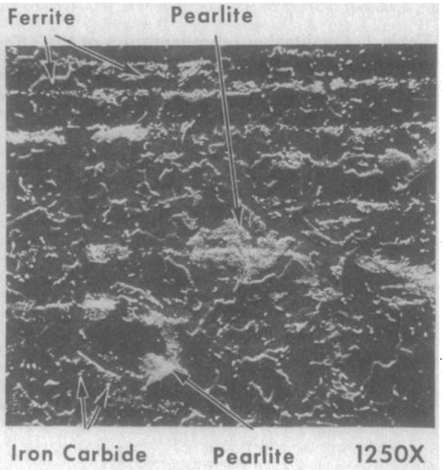


Fig. 1 (left). Scanning electron micrograph of plain carbon steel. The microstructure consists of a ferrite matrix with grain boundary iron carbides and a small volume fraction of pearlite. Fig. 2 (right). Scanning electron micrograph of a high-strength, low-alloy steel. The microstructure consists of a fine-grained ferrite matrix, grain boundary iron carbides, and islands of pearlite.

ten one property is enhanced at the expense of another. Consequently, no single mechanical property is maximized but many properties are optimized so that the steel is rendered suitable for various applications.

Although HSLA steels have an excellent balance of mechanical properties, their formability or ductility was not enough for the manufacture of many press-formed components for automotive applications. For these parts, involving such modes of forming as stretching and drawing, the reduced formability of HSLA steels was a serious stumbling block. To make HSLA steels more widely acceptable to the automobile industry, it was imperative to improve their strength-formability balance. This problem was resolved in the mid-1970's by the development of dual-phase steels (*3*, *4*).

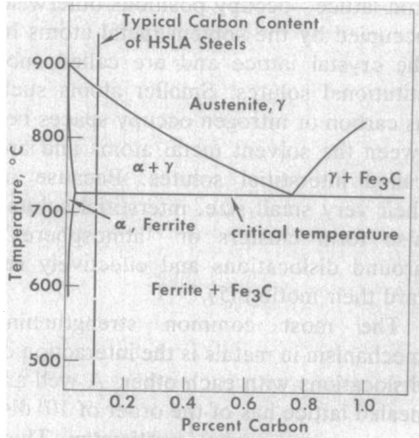
Steels described thus far have microstructures consisting of ferrite, pearlite, and iron carbides. The temperature range of stability of these phases is shown in the iron-carbon equilibrium phase diagram (Fig. 3). Ferrite or ferrite and cementite coexist in steel at temperatures below 723°C, the eutectoid or critical temperature (Fig. 3). For carbon contents less that 0.8 percent, cementite is not stable above the critical temperature, but austenite, or γ-iron, is stable to almost 1500°C.

When austenite is cooled slowly under equilibrium conditions, phase transformations occur at the temperatures indicated in Fig. 3. When cooled rapidly under nonequilibrium conditions, the austenite transforms to metastable phases, termed martensite and bainite, which cannot be shown on the equilibrium phase diagram. Martensite forms very rapidly by a shear mechanism when the steel is cooled quickly below a characteristic martensite start temperature. Martensite is far stronger than the other phases. Low-carbon martensite is needle- or lath-shaped. High-carbon martensite is usually twinned; the atomic arrangements of portions of the martensite grains are oriented in mirror-image fashion to accommodate strains introduced by the austenite-to-martensite transformation. Bainite is essentially an aggregate of ferrite grains with finely divided carbide particles and forms when the austenite is cooled and held just above the martensite start temperature.

Dual-phase steels are characterized by a ferrite microstructure with a uniform distribution of about 20 percent martensite by volume (Fig. 4). Their stress-strain curve is characteristically different (Fig. 5) from that of plain carbon or HSLA steels. All metals exhibit elastic behavior (a linear stress-strain relationship) at low strains and plastic behavior (a nonlinear stress-strain relationship) at higher strains. The transition from elastic to plastic behavior is continuous in dual-phase steels. Other steels exhibit yield point elongation or plastic deformation at a constant load before the steel work hardens and strength increases with increasing strain (Fig. 5). Yield strength is low in dual-phase steels compared to tensile strength, and strength increases rapidly as a function of strain as a result of a very high initial strain hardening rate. The relation of strength to uniform elongation (ductility) for plain carbon and HSLA steels can be represented by the same curve (Fig. 6). Similar data for dual-phase steels fall on a separate curve; these steels have higher ductility and are more formable than HSLA steels of equivalent tensile strength (Fig. 6).

Currently, dual-phase steels are produced with tensile strengths in the range 485 to 850 MPa. They have the same nominal chemistry range as HSLA steels and are produced by proper annealing of certain HSLA steels to convert their ferrite-pearlite microstructure to a ferrite-martensite one. Dual-phase steels are generally regarded as an extension of the broad family of HSLA steels, although dual-phase microstructures can also be produced using steel with no microalloying additions. HSLA and dual-phase steels are now produced and used commercially. Their balance of mechanical properties combined with the superior formability of dual-phase steels has substantially increased the utilization potential of high-strength steels in consumer products.

In this article I discuss some steel-strengthening mechanisms and present a brief treatise on how small amounts of alloying elements and controlled process-



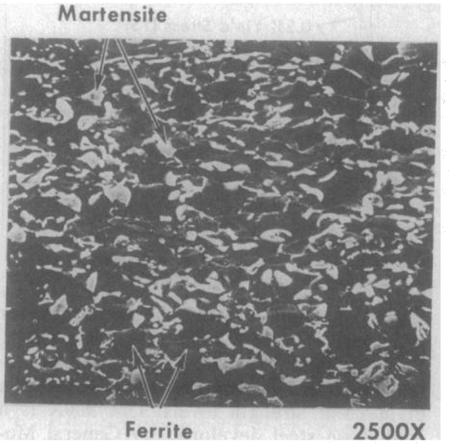


Fig. 3 (left). Schematic representation of a portion of the iron-carbon phase diagram. Fig. 4 (right). Scanning electron micrograph of a dual-phase steel. The microstructure consists of a fine-grained ferrite matrix with a uniform distribution of about 20 percent martensite by volume.

Table 1. Qualitative representation of the effects of various microstructural phenomena on the mechanical properties of HSLA steels. Beneficial effects are indicated by +, detrimental effects by −, and small or insignificant effects by 0 (*2*).

Property	Effect				
	Fine grain	Pearlite	Solid solution	Precipitation	Inclusion shape control
Strength	+	+	+	+	0
Formability	+/0	−	−/0	−	+
Toughness	+	−	−	−	+
Fatigue resistance	+	+	+	+	+/0
Weldability	0	−	−	−/0	+/0
Cost	0	0	−	−/0	−

ing can substantially improve mechanical properties of steels similar in chemistry to plain carbon steel. The contributions of the various strengthening mechanisms in ferrite-pearlite and ferrite-martensite high-strength steels are contrasted, and our present understanding of the mechanisms involved in transforming the former to the latter microstructure is discussed.

Strengthening Mechanisms

Plastic deformation occurs in polycrystalline metals by movement of dislocations on certain crystallographic slip planes within each crystal or grain. A dislocation is an extra half-plane of atoms in an otherwise periodic lattice; it is referred to as a linear defect and can move in the lattice under an applied stress. Strength can be increased by retarding dislocation motion. This is usually done by increasing the number of obstacles—precipitates, substitutional and interstitial solute atoms, grain boundaries, or other dislocations—in the path of the moving dislocations (*5*).

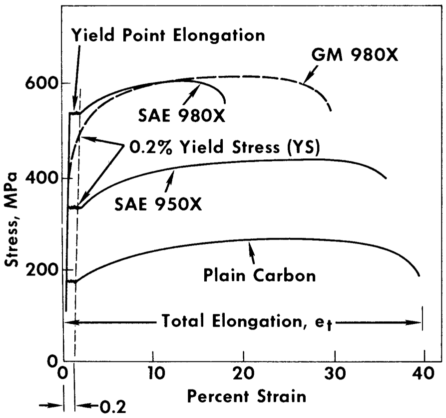


Fig. 5. Schematic stress-strain curves for plain carbon, HSLA, and dual-phase steels. SAE 950X and 980X are Society of Automotive Engineers designations for HSLA steels of different strength levels. GM 980X is a dual-phase steel developed by General Motors. GM 980X is more ductile than SAE 980X, although both steels have similar tensile strengths.

A uniform distribution of very small precipitates such as carbides, nitrides, or borides in a crystal lattice effectively retards dislocation motion. This contribution is enhanced when there is geometric matching, or coherence, between atoms of the precipitate and the matrix lattices. The volume, size, distribution, and coherence of precipitates can be controlled by steel chemistry and processing parameters. Moving dislocations are stopped when they run into precipitates, and the dislocation lines usually bend to avoid the precipitate and attempt to continue motion by a process called cross-slip (*5*). This involves slip on another set of crystallographic planes. Only a limited number of such planes are available, however, and dislocation motion is finally stopped; a much higher stress is required for further motion, resulting in stronger steel.

Strengthening by solute atoms is the result of elastic strains due to the difference in size between solute and solvent atoms and to solute-dislocation interactions. Solute atoms that are similar in size to atoms of the solvent metal—for example, manganese or silicon in an iron lattice—occupy positions otherwise occupied by the solvent metal atoms in the crystal lattice and are called substitutional solutes. Smaller atoms such as carbon or nitrogen occupy spaces between the solvent metal atoms and are called interstitial solutes. Because of their very small size, interstitial solutes also form clusters or "atmospheres" around dislocations and effectively retard their motion (*5*).

The most common strengthening mechanism in metals is the interaction of dislocations with each other. A well-annealed lattice has of the order of 10^6 dislocations per square centimeter. Their number increases with increasing strain, as do the interactions between them (*5*). However, because of the large number of dislocations, it is difficult to specify group behavior in a simple mathematical way or to calculate the exact strengthening contribution of the interactions.

Size of the crystal grains in the polycrystalline metal also has a significant effect on mechanical properties. Steel strength can be increased by decreasing grain size and thereby decreasing the distance available for free dislocation travel. The boundaries between grains are regions of lattice irregularity; dislocations can move relatively freely through the crystal but are impeded at the boundaries. Therefore, the smaller the effective size of the grains, the shorter the distance available for free dislocation travel.

Yield strength is more dependent on grain size than tensile strength. The beneficial effect of ferrite grain refinement on yield strength and on the temperature at which steels undergo the ductile-to-brittle transition was demonstrated in the early 1950's. Hall (*6*) and Petch (*7*) reported independently that a simple empirical relationship existed between yield strength and grain size, namely

$$\sigma_y = \sigma_i + \frac{k}{\sqrt{d}} \tag{1}$$

where σ_y is yield strength; σ_i is friction stress, a measure of the difficulty of moving dislocations through the lattice; k is the strength coefficient, a measure of the extent to which dislocations are piled up at obstacles; and d is ferrite grain size, in terms of equivalent diameter. This was one of the first demonstrations of a quan-

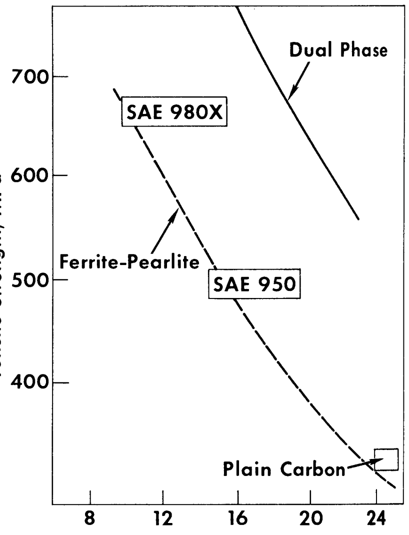


Fig. 6. Strength-ductility relationship of dual-phase steels compared with that for plain carbon and HSLA steels. The dual-phase steel curve is far above that for ferrite-pearlite steels (*13*).

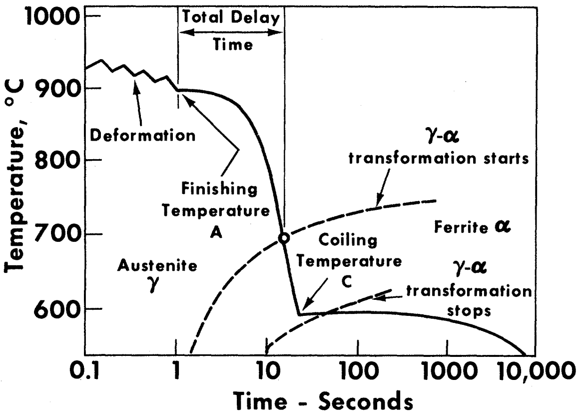


Fig. 8. Important processing parameters in a modern hot strip mill are indicated on a nominal cooling curve, which is superimposed on part of a schematic continuous-cooling diagram for a HSLA steel (*2*).

titative relationship between microstructure and mechanical properties.

Later, researchers used this relationship to explain and analyze quantitatively the effects of other structural factors involved in strengthening steels (*2*). This was done by expanding the friction stress to take into account solid solution strengthening by elements such as silicon and manganese and to allow for the volume fraction of pearlite in the microstructure and the effect of carbide and nitride precipitates, namely

$$\sigma_i = \text{function of } A + B\ (\%\ \text{pearlite}) + C\ (\%\ \text{Mn}) + D\ (\%\ \text{Si}) + \ldots + \Delta Y$$

where A is the friction stress in an unalloyed and unstrengthened ferrite matrix; B, C, D, . . . are empirical constants; and ΔY, the difference between σ_y measured and σ_y calculated, is assigned to the contribution of precipitates.

Examples of this empirical quantitative approach to structure-property relationships are shown in Fig. 7, where σ_y is plotted as a function of grain size. Each percent of manganese increases the strength about 48 MPa, each percent of silicon increases the strength about 150 MPa or more, and these contributions are additive. Individual contributions of various amounts of manganese, silicon, or pearlite to σ_y, and similar contributions due to precipitation strengthening related to vanadium and niobium additions, are also shown in Fig. 7. The various contributions to yield strength are superimposed schematically in Fig. 7 for a typical HSLA steel of 550-MPa yield strength and 650-MPa tensile strength. A typical stress-strain curve for one such steel (SAE 98OX) is shown in Fig. 5.

Similar relationships between other mechanical properties and various structural strengthening elements have been developed for HSLA steels. These effects are summarized qualitatively in Table 1. Grain refinement appears to be the most desirable method for improving several mechanical properties simultaneously. However, strengthening elements sometimes enhance one mechanical property at the expense of another. For example, microalloy precipitates increase steel strength but decrease formability and toughness. Such interactive relationships must therefore be considered in evaluating the contributions of various structural strengthening elements to each mechanical property. The relative contribution of each strengthening mechanism can be controlled through proper adjustment of chemical composition and processing variables.

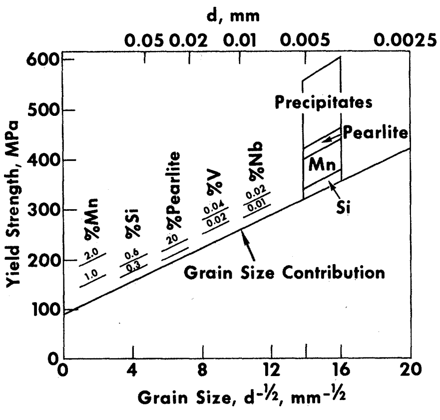


Fig. 7. Effects of grain size, solid solution strengtheners manganese and silicon, pearlite content, and precipitation strengthening elements vanadium and niobium on steel yield strength. A schematic representation of the origin of various strengthening contributions to yield strength of a typical HSLA steel with a tensile strength about 650 MPa is also shown (*2*).

Processing for Microstructural Control

Very briefly, the manufacture of steel consists of melting the steel and then pouring the molten metal into molds to solidify. These ingots are heated at elevated temperatures that are in the austenite region of the iron-carbon diagram (Fig. 3) to homogenize the steel, and then hot-rolled down in several stages to slabs 50 to 100 mm thick from an ingot 500 to 600 mm thick. The slabs are reheated as before and rolled down to a sheet of the desired thickness. As the sheet emerges from the rolling mill it is air- or force-cooled from the rolling temperature to a predetermined temperature and then coiled.

Effect of processing parameters. Steel microstructure and mechanical properties can be adjusted by controlling processing parameters during slab rolling. Some important parameters are shown qualitatively in Fig. 8; namely (i) rate and amount of deformation, particularly in the final rolls, (ii) finishing temperature, or the temperature at which hot-rolling is completed, (iii) time lapse between deformation and the beginning of the austenite-to-ferrite transformation, (iv) rate of cooling through the transformation range, and (v) coiling temperature. A portion of a schematic continuous-cooling diagram is also shown in Fig. 8. The continuous-cooling diagram defines contour lines for the start and finish of each phase transformation as a function of cooling time and gives an indication of the cooling rate required to develop the desired microstructure.

Work done during hot-rolling deforms the austenite grains and increases the strain energy in the lattice. With enough time and a sufficiently high temperature, the system tends to approach a low-energy equilibrium state by decreasing its high strain energy. This is manifest by recrystallization or nucleation of new grains having lower energy polygonal configurations, at the expense of deformed grains. The amount of deformation in the last rolling pass controls the number of lattice defects or dislocations present at the finishing temperature. A large number of defects provides in-

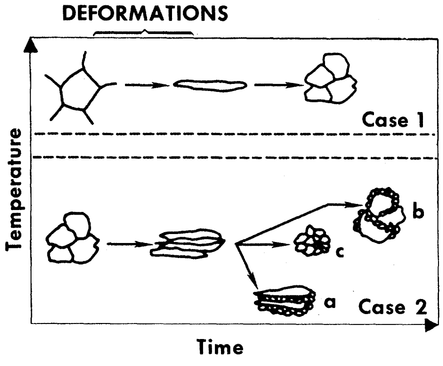


Fig. 9. Schematic representation of changes in austenite grains during controlled hot rolling (*10*).

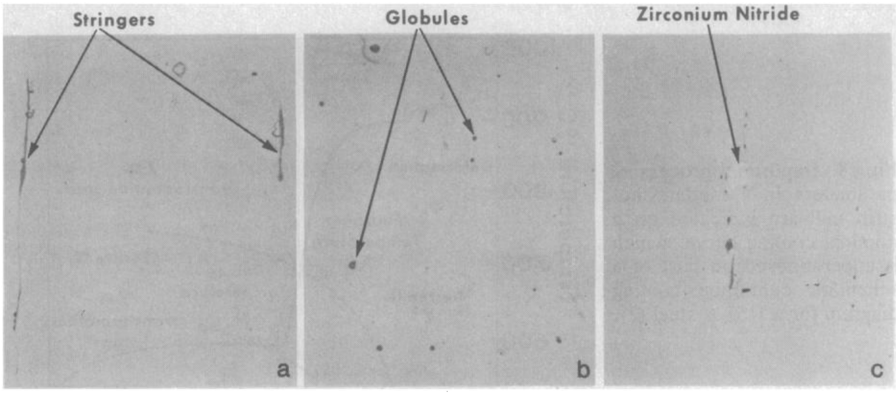


Fig. 10. Effect of inclusion shape control elements on shape of manganese sulfide inclusions. (a) Elongated stringers result from inadequate shape control. (b) Inclusions are globular, indicating adequate shape control. (c) Zirconium combines with nitrogen, forming zirconium nitride particles, and manganese sulfides are elongated (×400).

creased nucleation sites for recrystallization of austenite grains, which in turn results in a finer ferrite grain size on transformation of the austenite to ferrite.

Ferrite grain size can also be reduced by reducing the finishing temperature, because the two are directly related. In plain carbon steel, recrystallization occurs very rapidly during hot-rolling above 800°C and the recrystallized austenite grains grow to a fairly large size. The ferrite grain size that results on cooling through the austenite-ferrite transformation is then correspondingly large. On finish rolling below 800°C, recrystallization takes place more slowly and the austenite may only partially recrystallize. More ferrite nucleation sites are available in the partially recrystallized austenite, resulting in a finer ferrite grain size.

The cooling rate after hot-rolling governs the time available for transformation and grain growth. With higher cooling rates, less time is available for austenite grain growth, and this results in a smaller ferrite grain size on transformation. Furthermore, rapid cooling through the transformation range limits ferrite grain growth. Coiling temperature governs the amount of ferrite grain growth in the coil and is maintained as low as practicable to minimize such growth.

Effect of microalloying elements. The addition of a microalloying element such as vanadium, niobium, or titanium retards recrystallization and growth kinetics in the austenite and ferrite by several orders of magnitude compared to plain carbon steel, resulting in a finer ferrite grain size (*2*, *8*, *9*). The precise nature of this effect on recrystallization is still unresolved. Evidence has been presented both for and against the idea that the microalloyed elements must precipitate to inhibit recrystallization of the hot-rolled austenite (*1*).

The microalloy elements exist as carbide and nitride precipitates in the ingot and slabs. However, all but the largest precipitates dissolve in the steel during heating before hot rolling. During hot rolling, some strain-induced precipitation occurs at lattice defects and prior grain boundaries. These precipitates retard austenite recrystallization by pinning grain boundaries and subboundaries and limiting their motion (*8*). Also, dissolved microalloy solutes restrain grain boundaries and slow down their migration rate, retarding recrystallization (*9*). Similar mechanisms retard austenite grain growth following recrystallization. The high density of lattice defects and precipitates in the austenite provides an increased number of ferrite nucleation sites on cooling through the transformation, resulting in substantial grain refinement.

Higher finishing temperatures are generally necessary in microalloyed steels compared to plain carbon steel. When the finishing temperature is above 1000°C, austenite recrystallizes (*10*) and grains grow in size (Fig. 9, case 1), but are still smaller than would be expected in steels without microalloying additions. At lower finishing temperatures, say 1000°C, various morphologies may be developed, depending on steel composition and extent of prior hot-rolling history. Complete recrystallization may occur (Fig. 9, case 2c), generating a finer grain size than before (case 1). Partial recrystallization may occur at the grain boundaries of deformed austenite grains (case 2a) or grain growth may occur following complete recrystallization (case 2b). On transformation to ferrite, the austenite morphology developed by case 2c will yield the most uniform fine-grained ferrite.

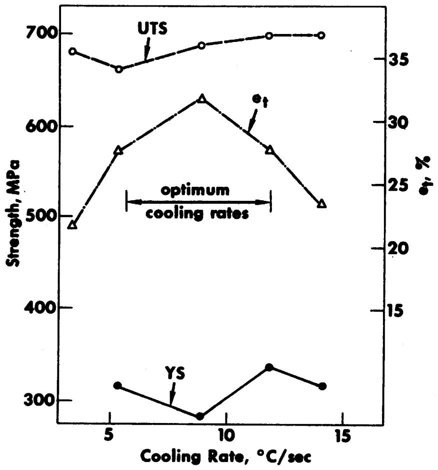


Fig. 11. Effect of cooling rate on mechanical properties of a dual-phase steel. Yield strength (*YS*) and ultimate tensile strength (*UTS*) are affected minimally in the cooling rate range evaluated, but ductility or total elongation (e_T) is substantially altered.

At still lower finishing temperatures (below 950°C), the deformed austenite will not have time to recrystallize and ferrite will nucleate from unrecrystallized austenite. The resulting ferrite grains are polygonal despite the distortion of the unrecrystallized grains, and the ferrite grain size becomes even finer than in case 2c.

As with plain carbon steel, rapid cooling between the finishing and coiling temperatures (points *A* and *C* in Fig. 8) inhibits austenite grain growth before transformation to ferrite and also lowers the transformation temperature, with further benefit to ferrite grain size. It also controls the intensity of precipitation strengthening and brings about a slight dislocation strengthening.

Controlled cooling. Rapid cooling is usually attained by pouring large volumes of cooling water over red-hot steel. On initial contact, the large difference between the steel and water temperatures promotes rapid heat transfer. However, an insulating layer of steam forms rapidly above the steel, lowering the heat transfer rate. Gradually, the steam jacket is penetrated and the steel and water again come in contact, resulting once again in a very rapid heat transfer. Efficient heat transfer can be attained by breaking down the insulating steam jacket as soon as it is formed.

Research has shown (*11*) that streams of water falling at low pressure on hot steel strip in a system called laminar flow cooling produce a rodlike flow of cooling water, which falls 1 to 2 meters before breaking into large droplets. Although to the eye this rodlike stream looks little different from any other low-pressure stream, it has sufficient energy to penetrate most effectively the protective steam blanket formed on the hot strip, achieve the required heat transfer and cooling rate, and develop the required microstructure and mechanical properties in HSLA steels. As discussed before, coiling temperature governs the amount of ferrite grain growth in the coil and is maintained as low as practicable to minimize such growth.

An added consequence of the hot-rolling process is its effect on the shape of nonmetallic inclusions in the steel. Since the latter influences steel strength and ductility, methods have evolved for controlling the shape of such inclusions by addition of suitable elements.

Inclusion shape control. Most steels, including plain carbon steel and HSLA steels, contain nonmetallic inclusions from the steelmaking process. These inclusions are globular in the ingot and consist primarily of manganese sulfide. At hot-rolling temperatures, the manganese sulfide inclusions are very plastic and are elongated into stringers in the direction of rolling (Fig. 10a). Consequently, steel ductility is inferior in the direction transverse to the rolling direction. To maintain their globular shape and prevent excessive stretching out of the sulfides during hot rolling it is necessary to alter their composition.

Zirconium or rare earth elements, such as cerium and lanthanum, form high-melting-point sulfides that are not readily deformable at hot-rolling temperatures. When these elements are added to molten steel, sulfur combines with them instead of manganese, forming rigid sulfides (*12*). After hot rolling, rare earth or zirconium sulfides remain globular (Fig. 10b). Most conventional HSLA steels are therefore treated with rare earth elements or zirconium for inclusion shape control to minimize anisotropy in mechanical properties. The use of zirconium is restricted, however, to steels that do not rely on nitride precipitation strengthening, because nitrogen competes with sulfur for the zirconium, forming zirconium nitrides (Fig. 10c). Besides tying up nitrogen instead of sulfur and failing to control the shape of inclusions and improve transverse ductility, zirconium nitride particles could contribute to overall reduction of ductility.

Dual-Phase Steels

Dual-phase steels are generally produced by continuous annealing (*4, 13*), but can also be produced as-rolled or directly off the hot mill (*14*). Continuous annealing involves moving a previously rolled strip of steel through a long furnace that is maintained at the desired annealing temperature. The required time at that temperature is ensured by controlling the rate at which the strip is fed through the furnace. In as-rolled dual-phase steels the desired microstructure is developed immediately after hot rolling by control of composition and cooling conditions. The manufacture of as-rolled dual-phase steels is similar to that of other HSLA steels, but the alloy composition is chosen so that 80 to 90 percent of the steel transforms to ferrite between the finish rolling and coiling temperatures and the cooling rate is controlled so that the remaining 10 to 20 percent austenite transforms to martensite in the coil.

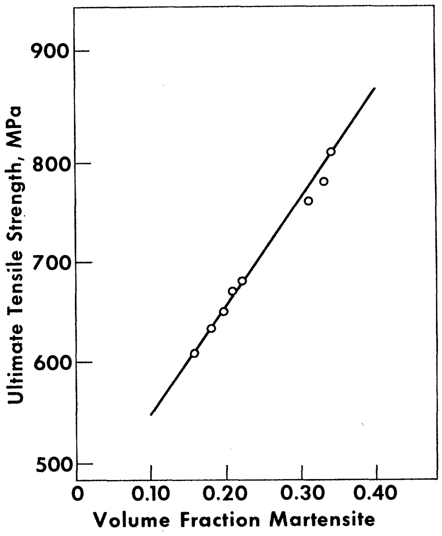


Fig. 12. Effect of volume fraction of martensite on tensile strength of a dual-phase steel. Tensile strength increases linearly with volume fraction of martensite (*15*).

Continuous-annealing parameters. Continuously annealed dual-phase steels are produced by heating HSLA steels of certain compositions intercritically, namely at temperatures in the ferrite plus austenite region of the iron-carbon phase diagram (Fig. 3), or supercritically, namely in the austenite region, and then cooling to room temperature. Processing parameters and other factors discussed with regard to HSLA steels do influence the mechanical properties of continuously annealed dual-phase steels. However, for each steel composition, the final microstructure and mechanical properties are governed principally by the continuous-annealing time and temperature, cooling rate after annealing, and coiling temperature.

For continuously annealed dual-phase steel, the composition is designed to have enough hardenability to produce about 20 percent martensite by volume at cooling rates possible on available processing equipment. Hardenability is defined as the ability of a steel to harden on rapid cooling. This occurs by the formation of martensite rather than ferrite and pearlite and can be enhanced by alloying with certain elements—for example, carbon or manganese. Steels with lower hardenability require rapid cooling, such as a water quench, whereas those with higher hardenability develop the required microstructure during air cooling. Lower cooling rates are generally preferred, because a larger number of lattice defects are quenched into the lattice at higher cooling rates and this can adversely affect ductility. Furthermore, optimum cooling rates exist for each steel composition for each heating time-temperature combination. For example, an e_T of more than 27 percent was reported for the steel in Fig. 11 for cooling rates between 5° and 12°C per second, e_T being less at other cooling rates. However, yield and tensile strength were minimally affected.

The continuous-annealing temperature determines the carbon content and strength of the martensite and hence influences the mechanical properties of the steel. In steels heated to lower intercritical annealing temperatures—for example, just above the critical temperature—the proportion of austenite formed is at a minimum and it has a high carbon content, because of the greater solubility of carbon in austenite than in ferrite (see Fig. 3). This austenite will transform to high-carbon twinned martensite on cooling. At higher annealing temperatures, the volume fraction of austenite is larger and it has a lower carbon content and hence lower hardenability. On cooling, some of the austenite will transform back to ferrite and the rest to low-carbon lath martensite. High-carbon martensite is stronger than low-carbon martensite, and hence their contributions to steel strength and ductility are different. Steels heated supercritically transform entirely to austenite. The volume fraction and carbon content of the martensite in these steels are different, and so are their contributions to mechanical properties.

The volume percent of martensite influences both strength and ductility of the steel. Tensile strength increases linearly (*15*) with volume percent of martensite (Fig. 12); ductility is inversely related to tensile strength (Fig. 6) and hence to volume percent of martensite.

After continuous annealing, the steel is cooled to below its martensite start temperature to ensure occurrence of the transformation, and is then coiled. Ferrite grain growth in the coil is not of much concern here; however, some tempering or softening of the martensite occurs as the coil cools to room temperature. Since martensite is a metastable phase it decomposes on reheating or tempering into more stable aggregations of ferrite and carbide, which have somewhat lower strength but improved ductility. Proper control of cooling parameters in the coil affords an additional means for controlling steel ductility.

The structure-property relationships. These relationships are quite complex in dual-phase steels. They are not fully understood but can be described in qualitative terms (*3, 4, 15*). Mechanical properties are influenced by steel composition and microstructure prior to continuous annealing. The fine grain size of the HSLA steel is inherited by the dual-phase steel, and Eq. 1 can be used to estimate the grain size contribution to yield strength. Ferrite grains do not appear to grow at continuous-annealing temperatures. Instead, some grain refinement seems to occur due to the martensite transformation. The strengthening contribution of precipitates in dual-phase steels may not be the same as those observed in HSLA steel because precipitate morphology is modified after continuous annealing. The relationship implied in Fig. 7 is altered, and the exact contribution of precipitates to the strength and ductility of dual-phase steel has yet to be isolated. However, substitutional strengthening contributions (Fig. 7) in the starting material appear to carry over into dual-phase steels.

Transformation mechanism. Before continuous annealing the steel microstructure consists of a parent ferrite phase with grain boundary iron carbides, small volume fractions of pearlite (Fig. 2), and a fine distribution of strengthening precipitates. On heating above the critical temperature the iron carbides go into solution, forming carbon-rich pools of austenite. On cooling back to room temperature part of the austenite transforms to martensite, part transforms back to ferrite, and a small fraction remains as austenite. Some precipitates are dissolved and large volumes of the ferrite are relatively free of them. For steels heated intercritically, some ferrite does not transform but the morphology of precipitates within this untransformed ferrite is usually altered (*4*).

These microstructural changes produce a stress-strain behavior (Fig. 5) that is quite different from that of plain carbon or HSLA steels. Dual-phase steels have no yield point elongation, a low yield strength, a high work hardening rate, and better ductility than HSLA steels of similar tensile strength. The low yield strength and absence of yield point elongation are attributed to the presence of numerous mobile dislocations in the ferrite generated by stresses imposed by the martensite transformation (*3*). Interactions between these dislocations and strain-induced transformation of the retained austenite to martensite at low strains contribute to the high work hardening rate. The high tensile strength is directly related to the volume percent of martensite (*15*). Mechanisms for the high total elongation and improved formability are quite complex; they are not fully understood but are generally explained as a consequence of the high work hardening rate.

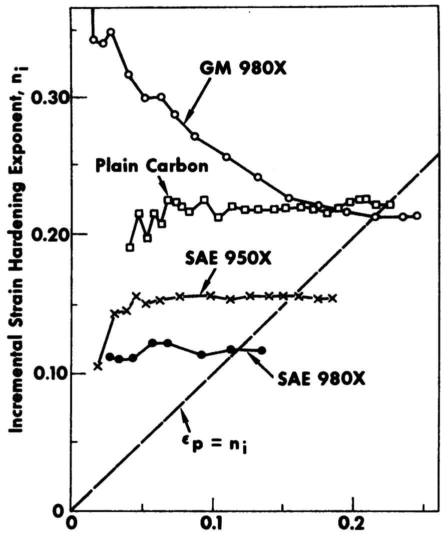


Fig. 13. Variation of incremental strain hardening rate, n_i, with increasing true plastic strain in various steels; n_i is relatively constant in the ferrite-pearlite steels but not in the dual-phase steel.

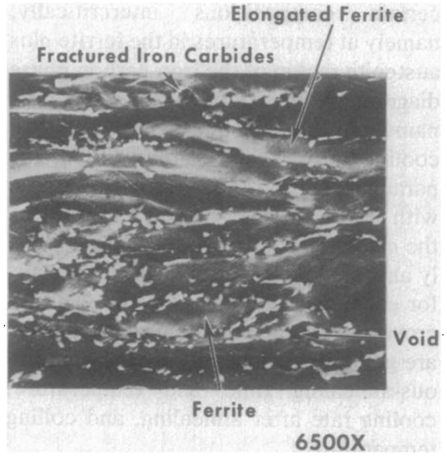


Fig. 14. Scanning electron micrograph of a ferrite-pearlite HSLA steel at fracture initiation. Ferrite grains are elongated, voids are created at ferrite-iron carbide interfaces, and some iron carbides are fractured.

Deformation mechanism. The deformation behavior of most metals, especially plain carbon steel, may be described (*16*) in terms of the simple empirical relationship between stress and strain, namely

$$\sigma = k\epsilon_p^{\,n} \qquad (2)$$

or

$$\log \sigma = \log k + n \log \epsilon_p \qquad (3)$$

where σ is the true stress, defined as the load at any instant divided by cross-sectional area of the specimen at that instant; k is the strength constant, or σ at $\epsilon_p = 1.0$; ϵ_p is the true plastic strain, defined as the change in length referred to the instantaneous length of the specimen; and n is the strain hardening exponent, a measure of the ability of the metal to distribute strain. When Eq. 2 is satisfied, a plot of log σ against log ϵ_p will be a straight line with slope n. Furthermore, it can be shown (*16*) that the true strain at maximum load, ϵ_u, also called the true uniform strain, will numerically equal n when Eq. 2 is satisfied.

Equation 2 is sometimes not satisfied, in which case a plot of Eq. 3 is a curved line (*17*). An incremental value of n, n_i, must then be calculated for each segment of this curve for small increments of strain, where

$$n_i = \frac{\log \sigma_j - \log \sigma_{j-1}}{\log \epsilon_{p(j)} - \log \epsilon_{p(j-1)}}$$

for $j \rightarrow 1$ to m, and m is the number of small segments on the stress-strain curve. The true uniform strain, ϵ_u, will now equal n_i at maximum load, and not the n value obtained for the full true stress-true strain curve (*18*).

A plot of n_i against ϵ_p will give an indication of the variation of the strain hardening behavior with increasing deformation in either type of steel. Equation 2 is satisfied in plain carbon and ferrite-pearlite HSLA steels, and a plot of n_i against ϵ_p is a straight line. This is not so for dual-phase steels (*19*). In the latter steels, n_i is very large at low strain but decreases progressively with increasing strain and is almost equal to that for plain carbon steels at maximum load (Fig. 13). The different variation of n_i with ϵ_p in the two steels gives an indication of the differences in their microstructural deformation characteristics.

The microstructure of dual-phase steels prior to continuous annealing consists of a parent ferrite phase with grain boundary iron carbides and some pearlite. When they are strained, plastic deformation takes place by strain hardening of the ferrite with practically no deformation of the iron carbides. Since deformation occurs predominantly in a single phase, n_l is constant and represents the strain hardening behavior of this phase alone (*19*).

Dual-phase steels consist of transformed and untransformed ferrite, martensite, and some retained austenite, all of which have different strengths. When dual-phase steel is strained, slip leading to deformation occurs first in the constituent with the lowest yield strength. When this constituent work hardens to the yield strength of the second constituent, slip begins to occur in it. This continues until all constituents are involved in the deformation process. The observed n_l at any particular strain is therefore the work hardening rate of one or several different constituents that may be undergoing deformation at that strain. Since deformation is shared consecutively by various constituents, strain is distributed more uniformly and failure is delayed. As a result, dual-phase steels have better formability than their parent HSLA steel (*19*).

After the ferrite is highly strained in the ferrite-pearlite steel, voids form preferentially at the ferrite-iron carbide interfaces (Fig. 14) and failure is initiated. In dual-phase steels, the martensite does not deform until the retained austenite transforms to martensite, and one or both ferrites, when present, are highly strained (*19*). Therefore voids leading to failure do not form until more extensive deformation has occurred and the martensite is also highly strained (Fig. 15). Hence failure occurs at much higher strains in dual-phase steels than in ferrite-pearlite steels of the same strength.

Attempts have been made to treat dual-phase steels as two-component composites, and their mechanical properties have been predicted with reasonable success (*20, 21*). However, further refinement of these models is necessary so that they may be used effectively in designing new steels with improved combinations of strength and ductility. Commercial utilization of all HSLA steels is growing rapidly and limitations in their mechanical properties are now being discovered. This has provided a new impetus to develop steels with modified plain carbon steel chemistry but with strength-ductility combinations that far surpass those of the available HSLA steels.

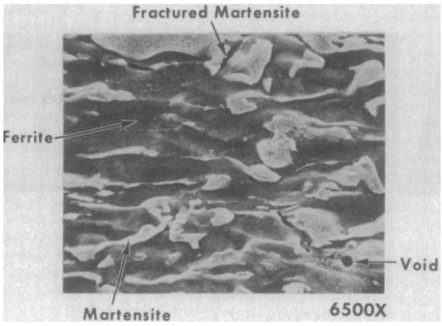


Fig. 15. Scanning electron micrograph of a dual-phase steel at fracture initiation. Both ferrite and martensite grains are elongated. Some martensite islands are fractured and voids are formed at the ferrite-martensite interfaces.

Conclusion

Slight modification of the plain carbon steel composition and close control of processing variables produce desirable properties in HSLA and dual-phase steels. Although an extension of the broad family of HSLA steels, dual-phase steels have different microstructures and mechanical properties. Their strength-ductility relationships are such that high-strength steel usage may be extended; greater weight reduction in steel products with complex shapes is now possible. The HSLA steels produced thus far have an optimized balance of mechanical properties that makes them suitable for various commercial products. However, in order to efficiently design components and conserve material and energy, the steel should be matched to the application. To accomplish this a more complete understanding of HSLA steels at the basic level is needed, and research in this field will undoubtedly lead to further steel developments.

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