Physical metallurgy of high-strength, low-alloy line-pipe and pipe-fitting steels

A. M. Sage

The effect of the characteristics of plate mills for production of plate and U–O–E pipe, and of continuous mills for production of strip and spiral pipe, is discussed in relation to ferrite–pearlite and two-phase steels. The influence of the cold work applied during pipemaking and expansion on the behaviour of each type of steel is considered, and the role of microalloying elements, individually and in combination, is discussed. The structure and properties achievable in normalized steel for pipe fittings are reviewed, and the effect of microalloys on weld properties is summarized. *MT/985*

 The author is with Highveld Steel and Vanadium Corporation Ltd, Westerham, Kent.

Unlike most other structures, line pipe (except when quenched and tempered in pipe form) is used in the cold-worked condition and the process of cold working the steel plate during pipe forming and expansion, when this is carried out, plays a critical part in determining the properties of the pipe. In some steels it lowers the strength of the pipe compared with the plate, and in others it increases the strength; in almost all steels it reduces the steel toughness. In pipe forming, the inner layers of the plate are deformed in compression, the deformation increasing from zero at the neutral axis to a maximum at the surface, while the outer part of the plate is deformed in tension.

During expansion, the inner layers of a pipe wall are deformed in tension and the outer layers receive further deformation in tension. In the flattening of a pipe section to provide testpieces, the inner layers are deformed in tension and the outer layers in compression. These deformations are summarized in Fig.1.

The properties of line pipes, like all structural steels, are controlled by the microstructure which in turn is determined by:

(i) the effects of the alloying elements on the kinetics of transformation and the formation of intermetallic compounds, i.e. the effects on the basic structure
(ii) the effect of secondary constituents on the structure, such as inclusions, which are in turn controlled by
(iii) steelmaking and casting processes and hence by the plant available for making and treating steel
(iv) the properties required by the pipeline engineer.

These groups of practical parameters determine the selection of steel compositions and steel treatments and, when taken together with our theoretical understanding of the relationship between metal structure and properties, can be considered the 'weft and warp' of steel development, as illustrated schematically in Fig.2. There is indeed a wide variety of steel patterns available and in use today for pipe made in mills having different characteristics and being supplied to many different markets.

In this paper, we are concerned primarily with the effects of alloying elements and processing of plates in relation to our theoretical understanding and the engineer's demands, but any decisions on steel composition selection must take all selection criteria into account.

There are many ways in which pipeline steels can be classified, but the most important division should be made on the basis of rolling, in as much as there are fundamental differences between the rolling of strip on continuous mills as used for spiral pipe, on the one hand, and plate rolled on reversing mills which is made into pipe by the U–O process followed by expansion, on the other.

Three types of steel with different basic structures having profoundly different effects on the mechanisms of deformation and the relationship between plate and pipe properties are used in both process routes. These are:

(i) the traditional ferrite–pearlite steels, albeit with low or reduced pearlite content, which exhibit a discontinuous stress–strain curve in the tensile test
(ii) ferrite–pearlite steels containing transformation products (bainite or martensite) to increase the work-hardening coefficient
(iii) steels containing polygonal ferrite and a major portion of a second phase, such as low-carbon bainite (sometimes called acicular ferrite) or islands of martensite and retained austenite which exhibit a continuous stress–strain curve in the tensile test.

The yield strength is determined by several parameters which can be expressed in a modification of the Hall–Petch[1,2] relationship for simple carbon steel (where only grain size is involved) such as that proposed by Morrison and Chapman[3]:

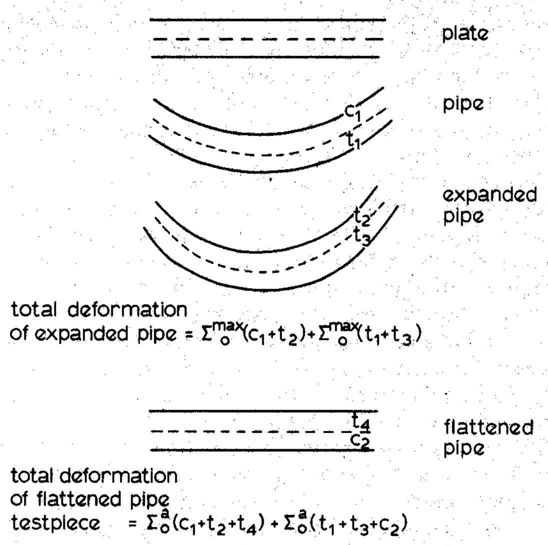
$$\sigma_y = \sigma_l + \sigma_{ss} + \sigma_{ppt} + \sigma_{disl} + \sigma_{text} + K_y d^{-1/2}$$

where

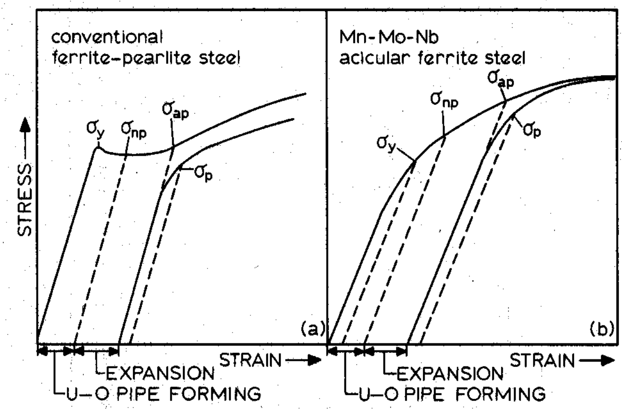
σ_y = yield strength
σ_l = lattice friction stress
σ_{ss} = solid-solution strengthening
σ_{ppt} = precipitation strengthening
σ_{disl} = dislocation strengthening
σ_{text} = texture strengthening
d = ferrite grain diameter
K_y = constant

As mentioned above, pipelines are structures operating

This paper was presented at the international conference 'Steels for line pipe and pipeline fittings', organized by The Metals Society and co-sponsored by the Welding Institute, and held in London on 21–23 October 1981. The conference was held as part of the celebrations of the 150th anniversary of the discovery of vanadium. The purpose of the conference was to bring together pipeline designers and constructors, and those from the steelmaking and welding industries, to examine trends in design, the potential of new steels, and new welding processes and testing methods. The proceedings of the conference, including discussions, are available as a proceedings volume.



1 Deformation in converting plate to expanded pipe

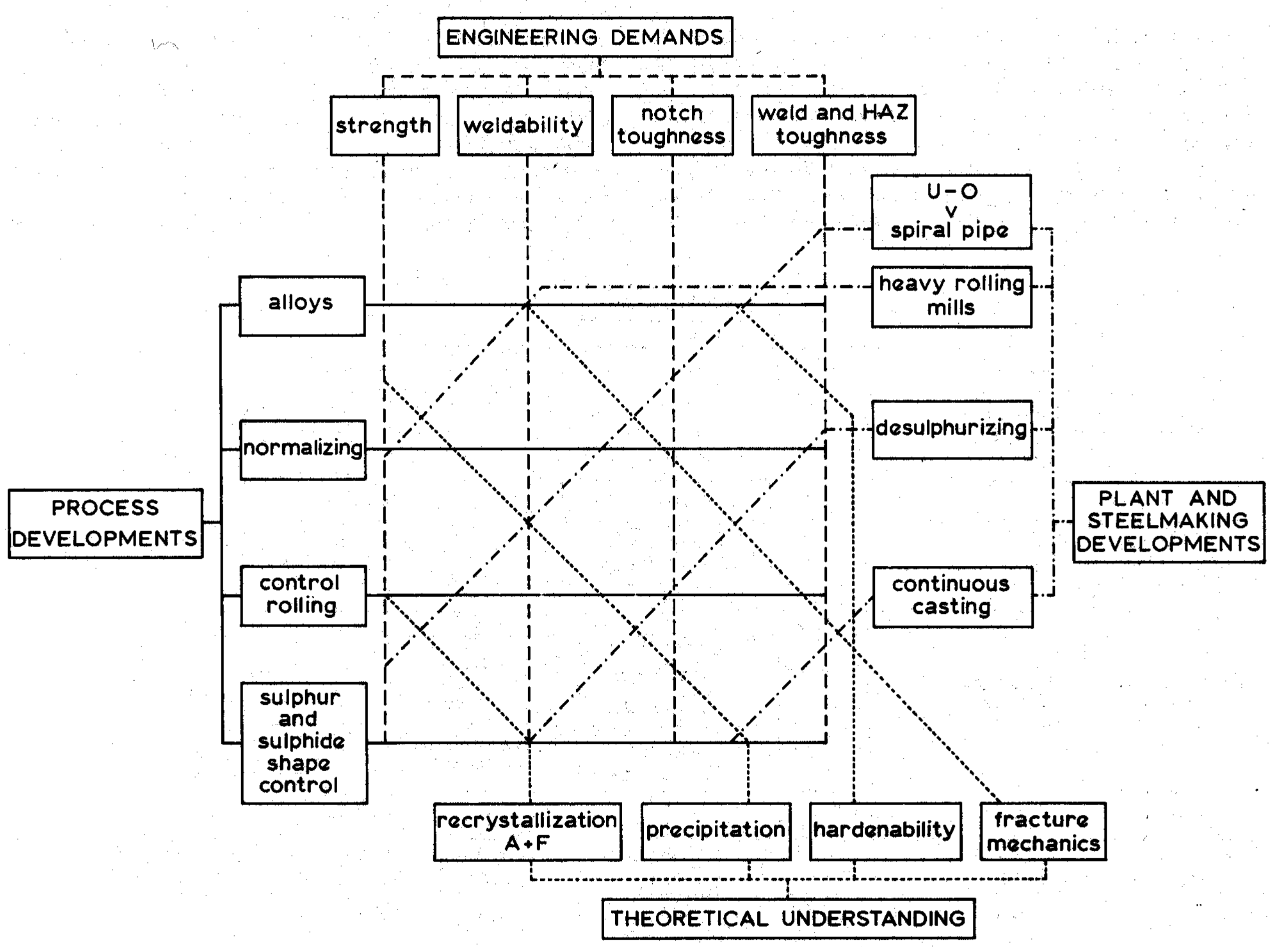


a ferrite–pearlite steel (type (i)); *b* steel containing second phase (type (ii))

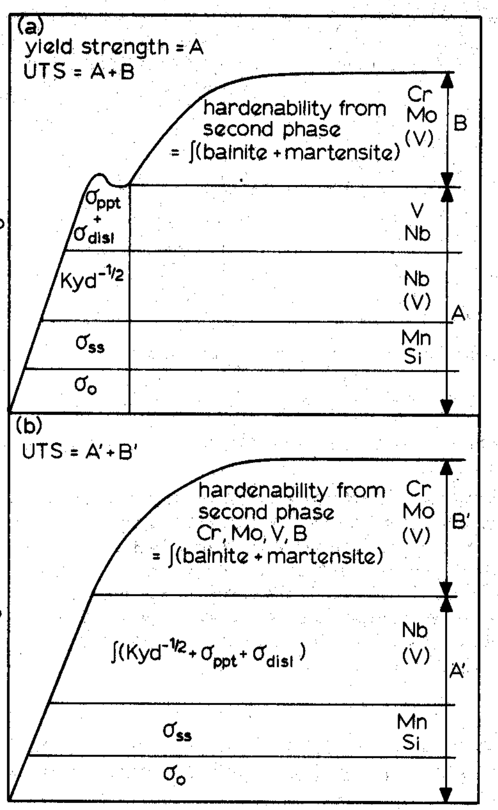
3 Schematic representation of how yielding behaviour of line-pipe steels affects yield strength of pipe, not to scale; σ_y yield strength of plate, σ_{np} 'actual' yield strength of non-expanded pipe, σ_{ap} 'actual' yield strength of expanded pipe, σ_p yield strength of expanded pipe (flattened tensile specimen), loss in yield strength due to Bauschinger effect $=\sigma_{ap}-\sigma_p$; after Ref.5

with steels in the cold-worked condition, and the steels go through at least one reversal of stress during the pipe forming, expansion, and test-preparation operations. Orowan[4] showed that strong barriers to dislocations, such as precipitates, created back-stresses due to dislocation pile-ups which could be wiped out by reverse straining, leading to permanent softening. Wilson[5] further showed that permanent softening increased with increasing volume fraction of precipitate and, at a constant volume fraction, increased with decreasing particle size. It is therefore common experience with the ferrite–pearlite steels described above that some loss in strength occurs between plate and pipe, a phenomenon first noticed by Bauschinger in 1881.[6]

The loss in strength for controlled-rolled steels which rely mainly on grain refinement (i.e. where $\sigma_{ppt}+\sigma_{disl}=0$) is small, and the phenomenon is therefore of little importance in most pipe made to X42 specifications.



2 'Weft and warp' of HSLA steel development



a yield strength and UTS of ferrite–pearlite steels (type (i)); *b* UTS of steels containing second phase (type (ii))

4 Parameters contributing to strength of line-pipe steels

In steels of type (i), where a significant part of the yield strength comes from $\sigma_{ppt}+\sigma_{disl}$, the loss in strength can be significant (as shown in Fig.3*a*) and the steel must be designed with a sufficient contribution from precipitation and dislocations to allow for this.

In steels of type (ii), this loss in strength due to permanent softening of precipitation/dislocation-strengthened steels can be partially offset by increasing the work-hardening characteristic of the steel, which increases the tensile strength of the deformed material (i.e. of the pipe).

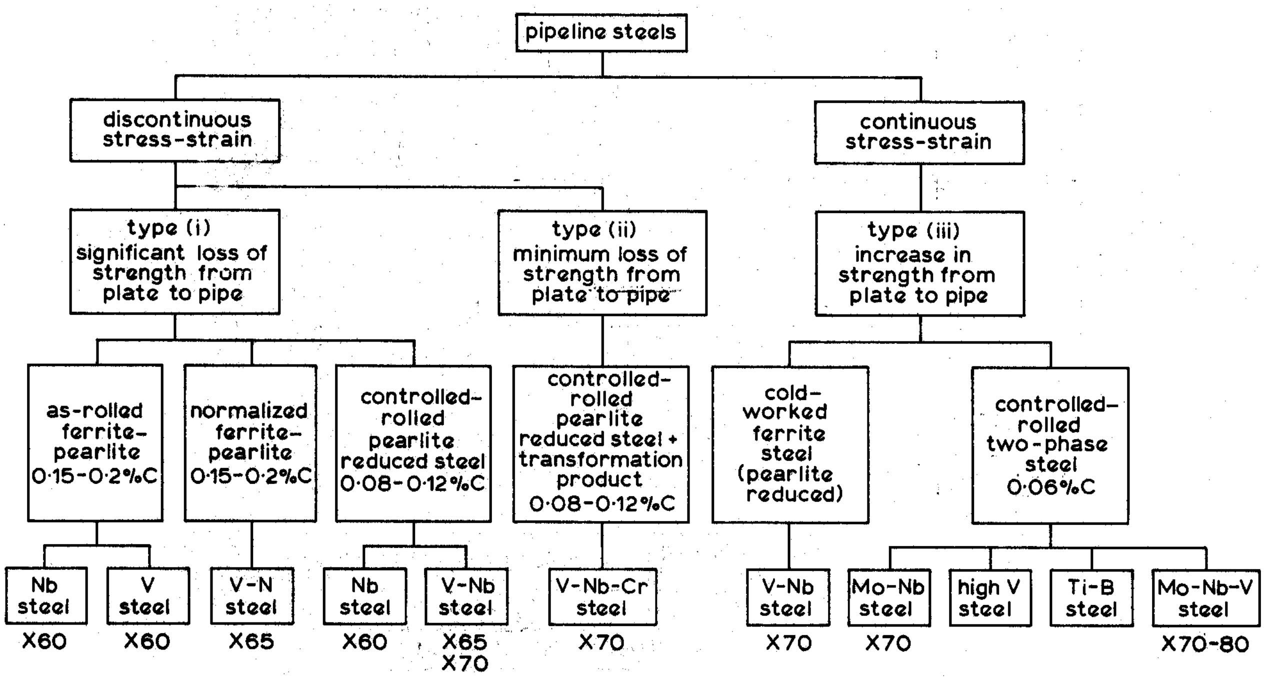
A schematic representation of the effect of the parameters of the Morrison–Chapman equation on yield strength, and of these and work hardening on the tensile strength, is shown in Fig.4*a*.

In steels of type (iii), the hard areas of the second phase under stress introduce a high density of dislocations into the surrounding ferrite. The work-hardening coefficient is therefore high, deformation takes place below the yield strength of the plate, and a continuous stress–strain curve results. Nevertheless, the parameters controlling yield strength influence the stress at which the initial deformation commences, as shown in Fig.4*b*, and the pipe strength can therefore be greater than the plate strength, as indicated schematically in Fig.3*b*.

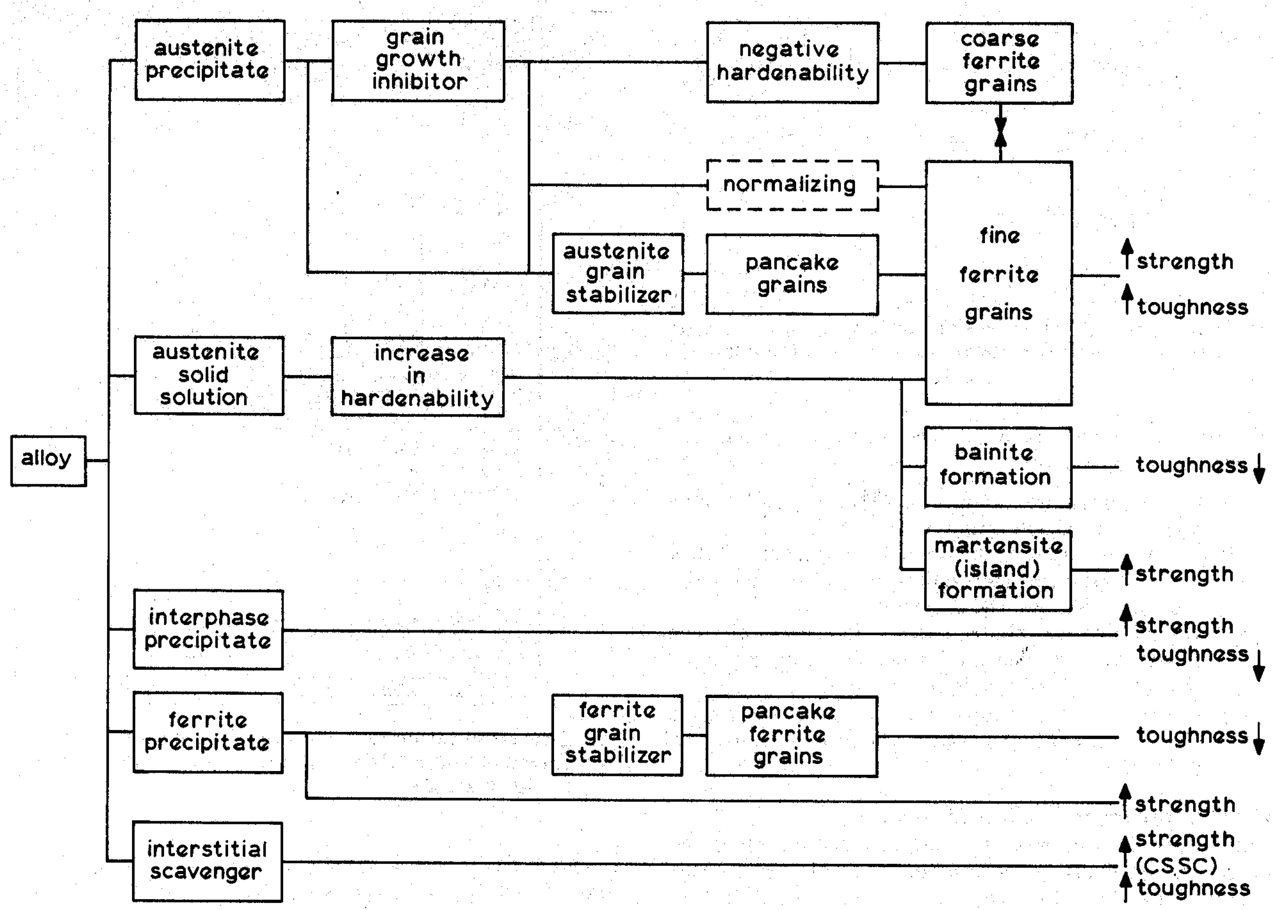
The steel types and structures are summarized in Fig.5 in which the most common composition type for each structure–performance type is indicated. Normalized steels have been included for completeness, but these have been superseded by the lower-carbon, more weldable controlled-rolled steels.

The microalloying elements (Ti, Al, Nb, V, and Mo) play a critical part in determining the structure and properties of pipeline steels, and their effects are monitored by rolling schedules. The varying effects which these alloys can have on structure, and hence on properties, are summarized in Fig.6, and the most common effects of each are indicated in Figs.7–11. Most high-strength steels for pipelines are given some form of controlled rolling or thermomechanical treatment in order to develop the structure required to achieve the desired properties. These functions of the microalloying elements result partly from their effect in solid solution, but mostly from their effects as precipitates. There is little evidence to indicate that their influence is related to precipitate morphology or their composition *per se*. Their effect is therefore a function of their size and the temperature at which they exist in relation to the transformation temperature of steel (the dynamic transformation temperature in the case of as-rolled steels). The temperature at which the most common microalloy compounds exist in relation to the transformation temperature is illustrated in Fig.12.

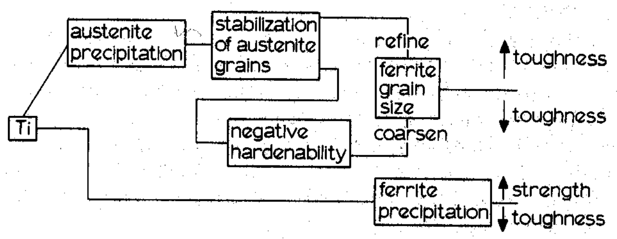
The temperature at which precipitates form and their



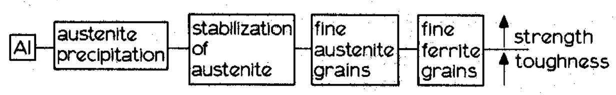
5 Summary of pipe steels in production for high-strength steel pipelines in Europe, the Americas, and Japan



6 Possible effects of microalloys on transformation structure and properties of pipeline steels



7 Effect of Ti on transformation structure and properties



8 Effect of Al on transformation structure and properties

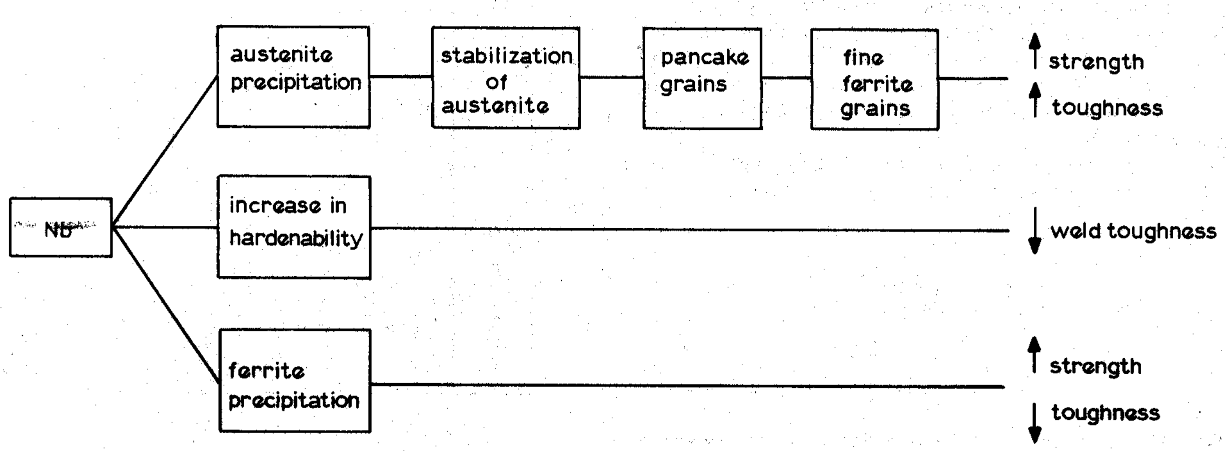
rate of growth are influenced by the rate of cooling (and hence plate thickness and any accelerated cooling given to the steel) and the presence of alloys in solid solution such as Cr, Cu, and Mo, all of which tend to suppress precipitation and therefore refine the precipitates.

The roles of individual microalloys in plate required for U–O processed pipe and in strip used for spiral pipe will be discussed separately.

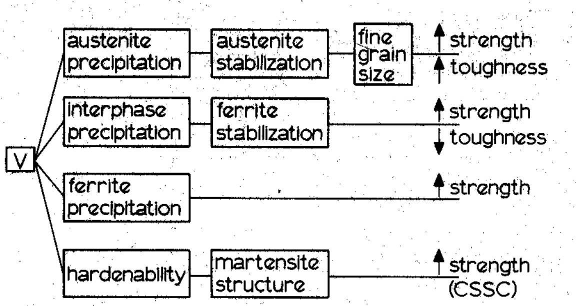
Plate and U–O pipe

STRUCTURAL CHANGES DURING CONTROLLED ROLLING

The rolling process for plate can be considered as taking place in four stages, although only two or three stages are



9 Effect of Nb on transformation structure and properties



10 Effect of V on transformation structure and properties; CSSC = continuous stress–strain curve

involved in many practices. These stages, which are illustrated in Fig.13, are:

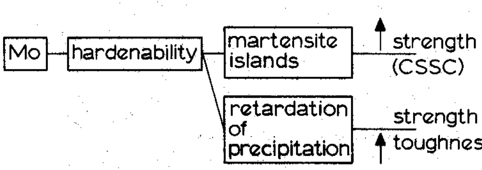
(i) high-temperature rolling of austenite when it may recrystallize if deformation is sufficient
(ii) rolling of austenite between a temperature below which it does not recrystallize or is prevented from recrystallizing and the Ar_3 temperature
(iii) rolling between the Ar_3 and Ar_1 temperatures when ferrite may recover or recrystallize and when an irregular grain size may be produced
(iv) rolling at a low temperature when ferrite will be cold worked and will not recrystallize unless heavily deformed.

High-temperature rolling of austenite

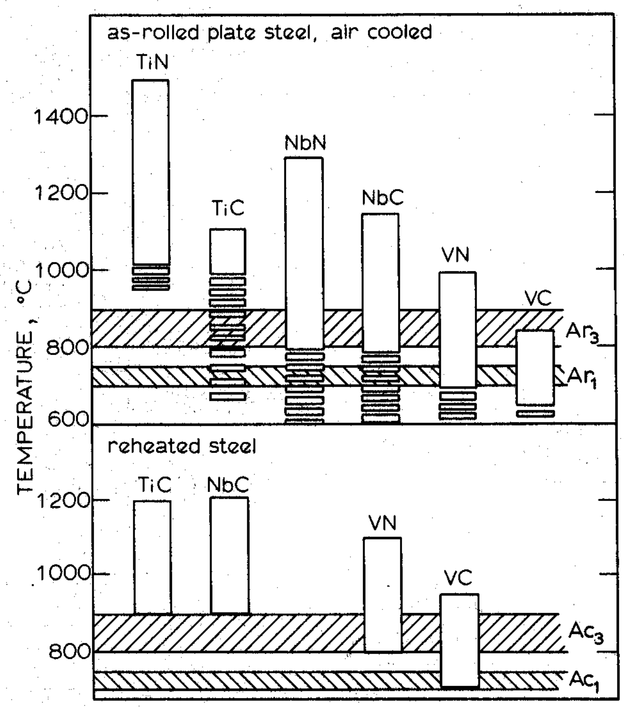
One of the most important parameters controlling the properties, particularly the toughness, of all steels is the austenite grain size at the end of the initial stage of rolling. This initial austenite grain size controls the grain size of the unrecrystallized austenite in the second stage and hence controls the ferrite grain size. The fineness of the ferrite contributes to the strength and to the toughness of all steels. Various procedures have been adopted, and others proposed, for refining the austenite grain size. The most commonly employed is a low reheating temperature because it has been demonstrated (by Irani *et al.*,[7] for example) that a low reheating temperature prevents austenite grain growth because NbC particles pin the grain boundaries. George *et al.*[8] have shown that TiN can have a similar effect. The lower reheating temperature also reduces the rate of grain-boundary movement and thus directly contributes to a fine austenite grain size, as has been shown by Priestner *et al.*,[9] at temperatures of, say, 1150°C which pin grain boundaries. Recent work[10] has also shown that in V steels containing no Nb or Ti, low reheating temperatures give finer grain size and higher toughness, even though vanadium carbide and nitrides are supposed to be in solution at this temperature.

It was shown by Matsubara *et al.*[11] that, provided a minimum deformation is given during the subsequent first stage of rolling, austenite will recrystallize, and that the grain size at 1 000°C depends on the degree of reduction and the availability of precipitates at this temperature to pin grain boundaries.

This is, however, in conflict with the results obtained by Cuddy,[12] who has shown in laboratory experiments that



11 Effect of Mo on transformation structure and properties



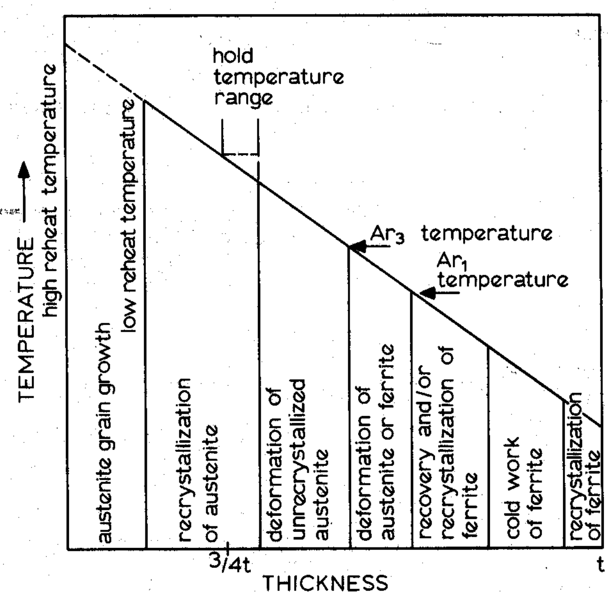
12 Estimated temperature ranges in which microalloy carbides and nitrides form on cooling and heating of HSLA steels

the grain size of a wide range of steels deformed at various temperatures is unaffected by initial austenite grain size, provided a minimum reduction of 70% is given.

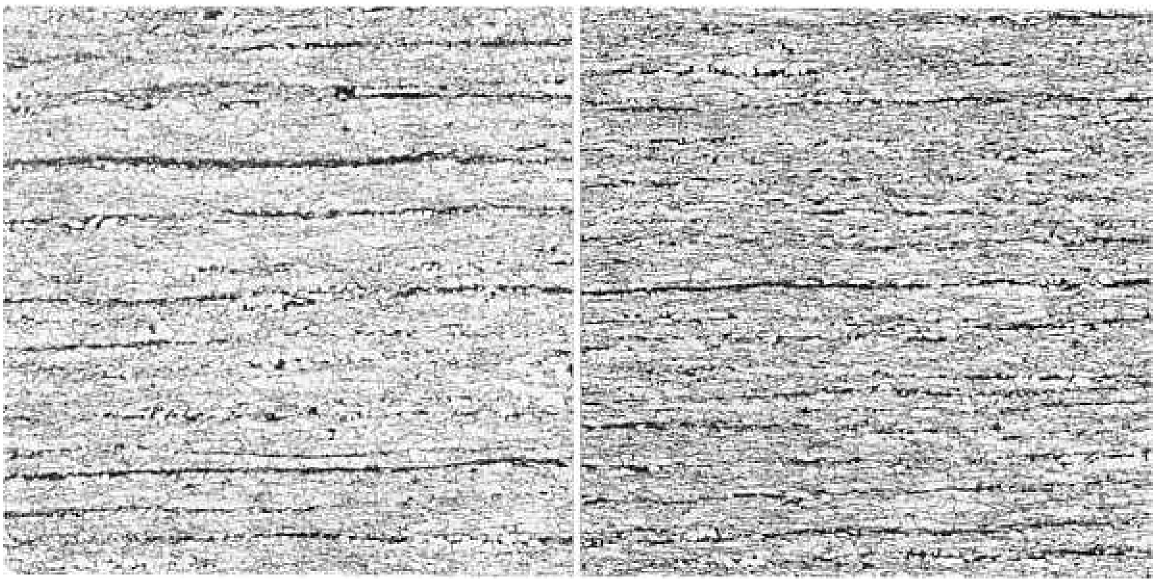
Low-temperature rolling of unrecrystallized austenite

In this stage, as first shown by Irani *et al.*,[7] the austenite grains are rolled into pancakes between the start rolling temperature and the Ar_3 temperature. During transformation, ferrite grains nucleate on both sides of the deformed grains and develop a grain size approximately half that of the minor axis of the pancaked austenite grains. (Some grains are also reported to nucleate within the austenite grains.)

It is generally considered that recrystallization of the austenite grains is prevented by precipitates which form



13 Stages in thermomechanical rolling of steel



a 0·007% N; *b* 0·012% N

14 Grain shape of controlled-rolled 20 mm thick 0·12V–N steel plates

above about 900°C, and NbC is the most commonly used for this purpose; sufficient precipitation is achieved with 0·03% Nb. It has also been suggested by Luton *et al.*[13] that the prevention of austenite recrystallization is caused by solute drag.

It has recently been shown by White and Owen[14] that the prevention of recrystallization can be achieved by VN in steels containing 0·2% V and 0·02% N. This is of little more than academic interest at the present time. Sage,[10] however, has shown that in a 0·45V–0·006N steel sufficient precipitation occurs to prevent austenite recrystallization, and there is no need to add niobium for grain-size control.

The extent of grain refinement achievable by this practice is determined by the amount of reduction given to the steel between the start rolling temperature (usually about 880°C) and the Ar_3 temperature. In most practices a reduction of about 75% in thickness is given during this stage of rolling, and it has been shown by Sellars[15] that greater reductions have little further grain-refining effect.

Rolling in the two-phase temperature range

Less is understood about rolling in this range, partly because the changes in structure are complicated by the fact that, while the first ferrite formed is being deformed and theoretically may or may not recrystallize, other grains of austenite are transforming and give rise to new undeformed equiaxed ferrite grains, and there is a danger that a structure of mixed grain size, detrimental to toughness, will be formed.

However, Sage,[16] in a limited study, has shown that in some steels rolled in this temperature range and at slightly lower temperatures, a limited amount of interphase precipitate is to be found and equiaxed ferrite can be produced, and that this is associated with increased toughness.

Recently, Bufalini and Aprile[17] have shown that improved toughness can be obtained in ferrite–pearlite V–Nb steels if they are rolled at temperatures just below the Ar_3 temperature. Unfortunately, they did not study the structure of the steel. Sage[16] also indicated that the equiaxed structure occurs only in high V–N or V–Nb steels when the precipitation of VN or NbCN takes place in austenite, so that light interphase row precipitation is found. He observed that in a low-V steel in which there had been interphase precipitation, recrystallization had apparently not occurred and deformed ferrite was produced, and that this was associated with poorer toughness (*see* Fig.14).

Relatively little work has been published on the structure of steels rolled in the two-phase region and below the Ar_1 temperature, and certainly little is known about the effect of various precipitates on the recrystallization behaviour of ferrite. Greater knowledge of the phenomenon might enable improved control of line-pipe properties to be obtained.

Rolling at low temperatures (in the ferrite range)

In some rolling practices, such as that discussed by Ouchi *et al.*,[18] rolling has been continued down to 690°C. Little has been reported on the structure of steels rolled at such low temperatures, although it is known that ferrite–pearlite steels of similar composition sometimes give a continuous stress–strain curve, and sometimes a discontinuous curve, in the tensile test when rolled at these temperatures. Rolling in this temperature range cold works ferrite, which could explain the continuous stress–strain curves. It is also possible that recovery could occur or that heavy deformation at this temperature could result in recrystallization of the ferrite, which would account for the discontinuous curves.

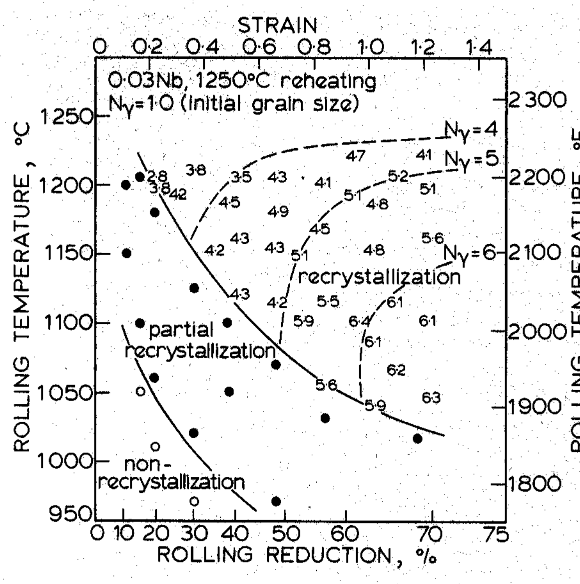
EFFECT OF MICROALLOYS ON CONTROLLED ROLLING OF FERRITE–PEARLITE PLATE STEELS FOR U–O PIPE PRODUCTION

Effect on ferrite

The microalloying elements Ti, Al, Nb, and V play important roles in determining the response of steels to the effects of the four stages of rolling, some or all of which are employed in thermomechanically treated steels. Each of the microalloying elements has separate, albeit overlapping, functions, and the role of some of them is considerably influenced by the level of nitrogen in the steel, and to a lesser extent by the carbon content.

The effect of the microalloying elements is determined by the temperature range over which their carbides and nitrides form, which in turn is controlled by the solubility of these compounds, their heats of formation, and the amount of nitrogen and carbon present in the steel.

Carbides and nitrides which form at a high temperature, and which do not or only partially enter solution at the reheating temperature, will prevent grain growth and thus



15 Austenitic recrystallization and resulting grain size in an Nb steel; after Ref.8

ensure a fine austenite grain size at the commencement of rolling. TiN is the most refractory microalloying compound and will, at all normal reheating temperatures, ensure a relatively fine starting grain size. NbC, certainly at a low reheating temperature, will also be undissolved and will have a similar effect.

During the first stage of rolling, austenite will recrystallize provided the total deformation exceeds a certain minimum at a given temperature (*see* Fig.15). If not inhibited by precipitates, the recrystallized grains will grow while the steel is above about 900°C. Precipitates which have remained from prior rolling, however, or precipitates formed from alloys which have entered solution and reprecipitated during the first stage of rolling, will inhibit grain growth and contribute to a fine austenite grain size at the end of the first stage. TiC, NbC or NbN, and AlN are the most important compounds operating in this way and in their presence, especially if a low reheating temperature is employed, a fine austenite grain size should be attained at the commencement of the second stage.

During the second stage of rolling, precipitates which have formed in the latter part of the first stage and which are therefore fine, and/or precipitates which form during the early part of the second stage, prevent recrystallization of the austenite, with the result that pancake grains with a high aspect ratio are formed and transform to fine, approximately equiaxed ferrite grains with a diameter of less than half the minor axes of the austenite grains. NbCN is the major compound affecting austenite in this manner, and sufficient NbCN is produced by the addition of 0·03% Nb. As mentioned above, it has been shown that in V steels if the V–N solubility product is high, VN will have the same effect. Compounds such as TiN which form at higher temperatures do not always appear to be effective in preventing the recrystallization of austenite, possibly because the precipitates grow to be too coarse. At the end of the second stage of rolling, therefore, if suitable precipitates have been present the structure consists of fine elongated austenite grains which, if no further working is carried out at lower temperatures, will form into fine equiaxed ferrite grains; the finer the austenite grains, the finer the ferrite grains.

The third and fourth stages of controlled rolling are not always involved. When, however, a steel is rolled below the Ar_3 temperature in the third stage of controlled rolling, ferrite is deformed and may recover or even recrystallize, and there is also evidence that this recovery is influenced by precipitation. In steels where no precipitation occurs in austenite, such as low-V–N steels, heavy interphase row precipitation of carbon-rich VCN occurs which may inhibit the recrystallization of ferrite, and a deformed ferrite structure results. In other steels, such as Ti and Nb steels or high V–N steels where precipitation occurs in austenite, light interphase row precipitation may take place, or random precipitates of VC or niobium carbonitrides may form at lower temperatures; it is thought that these do not inhibit recrystallization and that this may explain the recovered or recrystallized equiaxed grains which are observed. However, relatively little work has been done on this subject and further study is required.

In the fourth stage of rolling, where heavily deformed ferrite may recrystallize, the role of precipitates has not been considered.

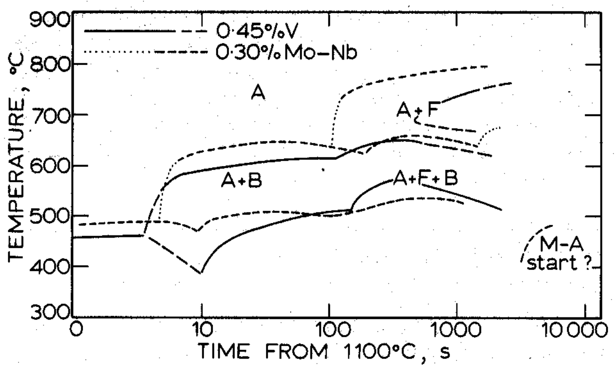
Effect on precipitation strengthening

Apart from the indirect effect of precipitates on grain size and grain morphology which influence the properties of the steel, precipitates can also have a direct effect on steel properties.

Precipitates which form in austenite, whether or not they influence austenite grain growth and recrystallization, usually grow to be too large to affect the properties of the steel directly, and only precipitates that occur in ferrite have a direct influence on steel properties. The ferrite precipitates occur in two forms: (i) as interphase row precipitates and (ii) as random precipitates in the ferrite forming at lower temperatures after the ferrite transformation has been completed.

All microalloys, except possibly aluminium, can form precipitates in ferrite, but naturally those which form carbides and nitrides in austenite are likely to be less effective ferrite strengtheners; the higher the temperature at which the compounds start to form, the less effective is the alloy likely to be, because the precipitation may be completed before the Ar_3 temperature is reached. The extent to which precipitation can be delayed so that some or all of the precipitation occurs in ferrite will depend on the concentration of the microalloy, the nitrogen and carbon contents of the steel, and the amount of solid-solution alloys Mn, Cr, and Mo present. It will also be influenced by the cooling rate of the ferrite, which in turn will be dependent on the plate thickness and any accelerated cooling which may be applied above the Ar_3 temperature. In plate steels, titanium and niobium, because of the high temperature at which their carbides and nitrides form, are in general less effective precipitates in ferrite than vanadium carbonitrides. Vanadium, if the concentrations of vanadium and nitrogen are not too high, may completely precipitate in ferrite either as interphase or random precipitates, and it is possible to move the precipitation of vanadium from austenite to ferrite by making small changes to the concentrations of vanadium and nitrogen, the cooling rate, and the amount of solid-solution alloys present, a phenomenon which is more important in normalized steel, as discussed below.

When more than one microalloy is present, the alloy forming carbide and nitride at higher temperatures can influence the form of the alloy precipitation at lower temperatures. This is particularly so when one of the alloys is vanadium. The presence of Ti, Al, or Nb, for example, can remove nitrogen from solution (forming large precipitates which have no effect on strength), and this prevents the precipitation of VN which might otherwise have formed in austenite. The presence of these alloys can therefore force more of the vanadium to precipitate in ferrite, thus increasing its effectiveness as a precipitate strengthener, but this has to be balanced by the fact that all the VCN may not precipitate in ferrite, i.e. some may stay in solution but can be precipitated on tempering.



16 Comparison of CCT curves for 0·45%V and 0·30Mo–Nb steels

There is also evidence[19] that microalloying elements can coprecipitate and that the coprecipitates can have different effects on properties from an aggregate of separate precipitates of the two alloys.

The effect of precipitation on strength is taken into account by the Morrison–Chapman formula for yield strength discussed above.

Formulae such as that given below[20] have also been derived for the Charpy-impact transition temperature:

$$T_1(°C) = -19 + 44\%Si + 700(\%N_f) - 11{\cdot}5d^{-1/2} + 2{\cdot}2(\%P)$$

where N_f is free nitrogen and P is pearlite.

There is, however, evidence to suggest that some precipitate forms have less effect on toughness than others. Sage,[16] for example, has found evidence that by increasing the nitrogen from 0·001 to 0·020% in a 0·12C–1·4Mn–0·18V steel, the precipitate is changed from mostly dense row precipitates to more random precipitates, with a consequent decrease in transition temperature of 80 K.

Effect on texture

The effect of ferrite texture is important only in steels rolled in the third and fourth stages, i.e. under conditions in which the ferrite is cold worked and does not recrystallize. Cold-rolled elongated ferrite grains produced during the second and third stages of rolling can be strongly orientated[21] and can give rise to splitting in the Charpy and drop-weight tests, which is sometimes considered to give a falsely high indication of toughness. The importance of splitting in Charpy or drop-weight fractured surfaces on the resistance of a pipe to fracture is a question for debate, though there is generally more acceptance of splitting in steel today than in the past.

TWO-PHASE STEELS

In the past few years, increasing interest has been shown in the two-phase steels described above.

When X70 steels were first required, attempts were made to upgrade the V–Nb controlled-rolled steels by increasing the strength through increased precipitation. When the Bauschinger effect was taken into account, however, this meant that plate yield strengths of over 550 MN m^{-2} were required, which was difficult to achieve and caused problems in pipe forming. Since then, however, improved control of rolling techniques has been developed to increase the strength by refining austenite and hence produce a fine ferrite grain size, and this reduces the loss in strength between plate and pipe.

Before this, however, the only alternative in plate steels for structural purposes was a quenched and tempered carbon steel. Facilities for quenching pipes were not available, and steels were therefore developed in which the hardenability (delay of ferrite formation) was used to produce a low-carbon bainite (sometimes called acicular ferrite) with a high work-hardening rate,[22] and steels containing 60% bainite and 40% polygonal ferrite were developed. However, Dabkowski and Speich[23] found that steel with islands of martensite in a ferrite matrix can be used to give the same strength. In these steels, particular advantage is taken of the effect of molybdenum on the pearlite hardenability, whereby pearlite formation is depressed and martensite islands are formed from retained austenite, in a matrix of soft ferrite.

Apart from the high tensile strength achievable by work hardening, these steels have another characteristic which is as important as their strength. The steel work hardens during the pipe-forming operation and the work hardening commences at a low stress, with the result that it is easier to ensure accurate shaping and good edge alignment in the U–O process, with consequently easier welding, fewer rejects, and higher productivity.

Furthermore, these steels have a lower carbon content than ferrite–pearlite steels, and this increases their weldability in spite of the higher proportions of solid-solution alloys they sometimes contain. Steels with various combinations of Mo, Nb, and V are being studied, and it is likely that they will be increasingly adopted as higher-strength (X80), thicker-walled pipes (for marine applications) are demanded by the engineer.

Until recently, two-phase steels were basically Mo–Nb steels in which the niobium provided grain refinement and some precipitation strengthening, and the Mo+Nb provided the hardenability (delay of ferrite and pearlite formation). More recently,[10] V steels have been developed in which the vanadium provides the grain refinement and hardenability.

Vanadium, unlike molybdenum, however, apparently has little effect on the ferrite transformation, but does delay the pearlite transformation and produces the martensite islands. Typical CCT curves for the Mo and V steels showing this difference are shown in Fig.16. Vanadium at the levels involved in these steels gives rise to some precipitation of VN in austenite, which is effective in preventing recrystallization of the austenite grains, and niobium is therefore not required.

It is well known that boron has a strong hardening effect on steel, and B steel (protected by titanium) is also being developed[11] for steels of this type.

Spiral pipes and strip steel

The conditions of rolling in a continuous mill making strip used for spiral pipe are different from those in a plate mill making plate for U–O pipe, and they place certain limitations on the steel, as well as providing some advantages.

While rolling practices similar to those used for plate can be used in the first stage of rolling in a reversing stand, the consecutive reductions in the five or six stands of the continuous mill mean that, unlike plate-mill rolling, there is no possibility of holding to a predetermined temperature before any specific pass, and the final rolling temperature is controlled by the entry temperature to the first stand. Moreover, the interstand time can be of the order of the time needed for austenite to recrystallize.

On the other hand, the quenching from about 800° to 650°C which the steel receives after leaving the last stand and before coiling gives the steel a fast cool through the transformation range. This gives rise to a fine ferrite grain size, and a niobium addition to give a fine grain size through prevention of recrystallization of the austenite is less necessary than in plate rolling.

The quick cooling of this steel from about 800° to 650°C

Table 1 Parameters of plate and strip mills controlling plate microstructures

Plate mill	Strip mill
Reheat temperature	Reheat temperature
Primary reduction	Primary reduction
Hold before final rolling	Hold before final rolling
Interpass time allows recrystallization of austenite	Interpass time approximately equal to time for recrystallization of austenite
Finish temperature controlled by interpass time and stand temperature at start of final rolling	Finish temperature controlled by temperature at start of final rolling
Slow cooling from 800° to 650°C	Fast cooling from 800° to 650°C
Slow cooling from 650°C to RT	Very slow cooling from 650°C to RT

at the end of rolling, apart from producing a fine ferrite grain size, retains alloys in supersaturated solution at this temperature so that they can precipitate during slow cooling of the steel from 650°C after coiling. This means that in ferrite–pearlite steels the precipitation of titanium and niobium in ferrite, and of vanadium in high-nitrogen steels, is more effective than in plate steel.

In two-phase steels the accelerated cooling not only leads to grain refinement and enhanced precipitation, but also contributes to the hardenability of the steels.

A comparison of the parameters affecting the structure of steels in plate and strip mills is given in Table 1.

Practical pipeline steels

It is not the purpose of this paper to discuss steel compositions in detail, but a summary of the steel types most commonly used or now being considered for the major types will be presented.

X60

Pipes for gas or oil are made to specifications based on API 5LX60, usually produced as controlled-rolled ferrite–pearlite 0·03% Nb steels, adequate strength being achieved from ferrite grain refinement and there being little loss of strength between plate and pipe.

X65

Pipes for gas or oil transport made to X65 specifications are made mostly as controlled-rolled 0·03Nb–0·06–0·09V steels, the strength being achieved by a combination of fine grain size and precipitation strengthening. Approximately 50% of the strength from precipitation is lost during pipemaking, and plates with a minimum of 470 MN m^{-2} are normally produced for this strength of pipe.

X70

Pipes for gas transport made to X70 specifications are made mostly as controlled-rolled 0·03Nb–0·06–0·09V steels with a special rolling practice involving a low reheating temperature and/or heavy primary rolling to produce an extra-fine austenite and hence a fine ferrite grain size. Some grades contain chromium to produce bainite in the structure and increase the work-hardening capacity. Where suitable rolling practices are not available or where heavier gauges are involved, however, Mo–Nb two-phase steels are used.

Marine pipes for gas or oil are usually of greater wall thickness than land lines and generally of lower strength. Current demands involve pipes of 25 mm wall thickness. The grain refinement achieved by controlled rolling tends to decrease with increasing plate thickness, and the precipitation strengthening tends to decrease owing to growth of the precipitate. Ferrite–pearlite V–Nb steels, however, with additions of molybdenum or copper, experience slower growth of precipitates and may find increasing applications in thick-walled pipes. Alternatively, Mo–Nb, 0·45% V, or Ti–B steels of the two-phase type will be considered.

Future developments

The steels for the production of line pipe are under constant development. On the one hand, the pipeline engineer is pressing for higher-strength pipe to decrease costs of gas transport, and thinner-walled pipes to reduce the cost of transport of the pipe to site. On the other hand, the steel producer is seeking ways to produce lower-cost steels by appropriate combinations of alloys and processing.

The land engineer is likely to require pipe to X75, X80, or X85 specifications, with wall thicknesses in the range 14–22 mm, while the marine-pipeline engineer is looking for steels of 25–30 mm wall thickness with X65 strength levels.

The limit in strength has probably been reached with ferrite–pearlite steels, and the higher-strength steels are likely to be of ferrite plus a harder phase. On the production side, as far as steels for present (and future) specifications are concerned, attempts to reduce costs of production are being made by the introduction of accelerated cooling of the steel during rolling of plate in the temperature range 850°–650°C (i.e. reproducing strip-mill cooling conditions in a plate mill).

Such cooling, which has been described by Ouchi *et al.*,[18] produces a fine grain size and offers the opportunity to avoid the use of niobium for grain refinement. It is also possible that, at least in the thinner sections, the rate of cooling could be sufficient to produce martensite islands without the use of alloys. In such cases, a low-vanadium or low-niobium steel should provide sufficient precipitation hardening to achieve high strength.

The processes for these developments are, however, still in the development stage and until their ability to give adequate uniformity of properties in plate free from hydrogen without distortion has been perfected, commercial development of these steels will not be seen.

Pipe-fitting steels

Steels for pipe fittings are supplied in plate thicknesses up to 100 mm or more, and mostly in thicknesses greater than those in which high strengths can be achieved by controlled rolling, although some suppliers offer controlled-rolled plates up to 40 mm thick.

Most steels used for pipe fitting, therefore, are supplied in the normalized condition or are normalized and tempered steels, though some quenched and tempered steels are also used.

The strength in these normalized steels, as with the controlled-rolled steel, is achieved by a combination of fine grain size and precipitation. In this case, however, the fine grain size is achieved by precipitates formed during rolling which do not redissolve at the normalizing temperature and pin the grain boundaries. The precipitation strength comes from precipitates of higher solubility which go into solution at the normalizing temperature and reprecipitate on cooling. Demicco *et al.*[24] have drawn attention to the fact that precipitates have to find suitable sites for their formation, such as dislocations produced during transformation or as a result of cold-worked ferrite. In normalized steels, however, there are relatively few sites except those produced during transformation, and most precipitation is therefore of the interphase-row type.

The most important compounds used to refine the grain size are AlN, VN, and NbCN, and the precipitation strength is achieved by carbon-rich VCN. It has recently been shown,[25] however, that the strength can be maintained in thicker plates by adding molybdenum, which suppresses the vanadium precipitation, making it finer and more effective. In some grades, tempering of the normalized steel is required to bring out the precipitate. Copper will probably have a similar effect. Unlike controlled-rolled steel, little strength comes from dislocations because the ferrite grains are almost dislocation free. The precipitation is therefore mostly interphase and, in order to achieve the desired strengths, carbon contents are higher than in controlled-rolled steels.

Welding of pipeline and pipe-fitting steels

In the submerged-arc and girth welding processes used for the seam and butt welding of pipelines, complete fusion of the steel occurs and the steel is diluted with a filler metal and cooled quickly. The weld metal is therefore a quickly cooled cast structure. The zone adjacent to the weld is heated to a high temperature and is also quickly cooled. In order to decrease the cost of welding, the engineer has developed high-speed processes which for seam welds involve high heat input and consequently slower cooling rates.

The structure of a weld metal usually consists of acicular ferrite with grain-boundary ferrite in varying proportions and ferrite aligned with martensite–austenite–carbide constituents, but the presence of microalloys, through their effects on hardenability, can modify these structures and, through the precipitation of carbonitrides, can increase strength and may lower the toughness of weld metal.

Vanadium, as indicated above, appears to suppress pearlite but not ferrite formation, and this behaviour is reflected in its effect on weld-metal structures. It therefore has very little effect on the proportion of grain-boundary ferrite, but does inhibit the growth of side plates, possibly through the effect of the VCN.

Niobium, on the other hand, which suppresses ferrite formation in rolled plates, also tends to reduce grain-boundary ferrite formation in weld metal and thus to increase toughness, but does not appear to inhibit growth of side plates, particularly at low levels of acicular ferrite in the structure.

Both elements can form precipitates, especially during the reheating of weld metal during a subsequent pass, and such precipitates tend to reduce toughness. From a practical point of view, about 0·02% Nb or 0·05% V in weld-metal deposits can usually be tolerated with no loss of toughness by using a correct choice of consumables. Higher levels of these microalloying elements tend to show reduced toughness, but amounts even greater than 0·05% Nb or 0·1% V in the deposit can give good toughness if the appropriate consumables are selected. Sage,[10] for example, showed that when welding a 0·45% V steel which gave over 0·25% V in the weld metal, good weld toughness was obtained. These effects of vanadium and niobium are discussed in more detail by Dolby.[26]

The HAZ is also an area prone to embrittlement owing to the grain growth caused by the high temperature it attains during welding. In the higher-carbon steels there is also a tendency for carbides such as vanadium to precipitate and add further to the embrittlement. In addition, nitrides tend to dissociate and under the high cooling rates to remain dissociated, leaving nitrogen in interstitial solid solution, which also adds to embrittlement.

Titanium and zirconium are the only elements which can overcome this because their nitrides are stable at high temperatures. They not only prevent free nitrogen being present, but also the TiN/ZrN particles inhibit grain growth and also nucleate new ferrite grains. A number of titanium-bearing steels have been developed to avoid HAZ embrittlement, especially where high heat input processes are used.

In developing such steels, however, account has to be taken of the effect of titanium on the formation of other alloying elements present. Titanium, if present in sufficient quantities in relation to the nitrogen content (because of its affinity for nitrogen, for example) can prevent vanadium precipitation in austenite and force it all to precipitate in ferrite, which for the reasons discussed above may or may not be desirable.

Acknowledgments

The author wishes to thank Dr W. Morrison, BSC Teesside Laboratories, and Dr R. Dolby, The Welding Institute, for their help in preparing this paper.

References

1. E. O. HALL: *Proc. Phys. Soc.*, 1951, **B64**, 742.
2. N. J. PETCH: *J. Iron Steel Inst.*, 1953, **174**, 25.
3. W. B. MORRISON and J. A. CHAPMAN: *Phil Trans. R. Soc.*, 1976, **A282**, 289.
4. E. OROWAN: in 'Internal strain and fatigue in metal', (ed. G. M. Rassweiler and W. L. Grabe), 59; 1959, Amsterdam, Elsevier.
5. D. V. WILSON: *Acta Metall.*, 1965, **13**, 807.
6. J. BAUSCHINGER: *Ziviling.*, 1881, **27**, 289.
7. J. J. IRANI, D. BURTON, J. D. JONES, and A. B. ROTHWELL: in 'Strong tough structural steels', 110; 1967, London, The Iron and Steel Institute.
8. T. J. GEORGE, G. BASHFORD, and J. K. MACDONALD: *J. Aust. Inst. Met.*, 1971, **16**, 36.
9. R. PRIESTNER, C. C. EARLEY, and J. H. RENDALL: *J. Iron Steel Inst.*, 1968, **206**, 1252.
10. A. M. SAGE: *Met. Technol.*, 1981, **8**, 94.
11. H. MATSUBARA, B. SAKAI, and T. ITAOKA: in 'Materials engineering in the Arctic', (ed. M. B. Ives), 190; 1976, Metals Park, Ohio, American Society for Metals.
12. L. J. CUDDY: Proc. Conf. 'Recrystallization and grain growth of multi-phase and particle-containing materials', Roskilde, Denmark, Sept. 1980, Risø National Laboratory, 317.
13. M. J. LUTON, R. DORVEL, and R. A. PETKOVIC: *Metall. Trans.*, 1980, **11A**, 411.
14. M. J. WHITE and W. S. OWEN: *Metall. Trans.*, 1980, **11A**, 597.
15. C. M. SELLARS: in 'Hot working and forming processes', 3; 1980, London, The Metals Society.
16. A. M. SAGE: in 'Hot working and forming processes', 119; 1980, London, The Metals Society.
17. P. BUFALINI and A. APRILE: in 'Sheet rolling', Vol. 2, 945; 1980, Tokyo, Iron and Steel Institute of Japan.
18. C. OUCHI, K. TSUKADA, and J. TANAKA: in Proc. Conf. 'Vanadium in high strength steel', Chicago, Ill., 1979, Vanitec, 37.
19. S. UEDA, M. ISHIKAWA, A. KAMADA, and N. OHASHI: *Trans. Iron Steel Inst. Jpn*, 1980, **20**, 759.
20. D. B. CLAY and D. B. McCUTCHEON: *Phil Trans. R. Soc.*, 1976, **A282**, 305.
21. E. INIGOSHI *et al.*: Proc. Conf. 'Crack propagation in pipeline', Newcastle upon Tyne, 1974, British Gas.
22. Y. E. SMITH, A. P. COLDREN, and R. L. CRYDERMAN: in 'Toward improved ductility and toughness', 119; 1971, Tokyo, Climax Molybdenum Development Company (Japan) Ltd.
23. D. S. DABKOWSKI and G. R. SPEICH: in Proc. Conf. 'Mechanical working and steel processing XV', Pittsburgh, Pa, 1977, Iron and Steel Society of AIME, 284.
24. D. R. DEMICCO, A. T. DAVENPORT, G. M. WAID, and R. A. MOSSER: in 'Steels for line pipe and pipeline fittings', 242; 1983, London, The Metals Society.
25. T. WADA and Y. E. SMITH: in 'Applications of materials for pressure vessels and piping', 31; 1979, New York, ASME.
26. R. E. DOLBY: in 'Steels for line pipe and pipeline fittings', 302; 1983, London, The Metals Society.