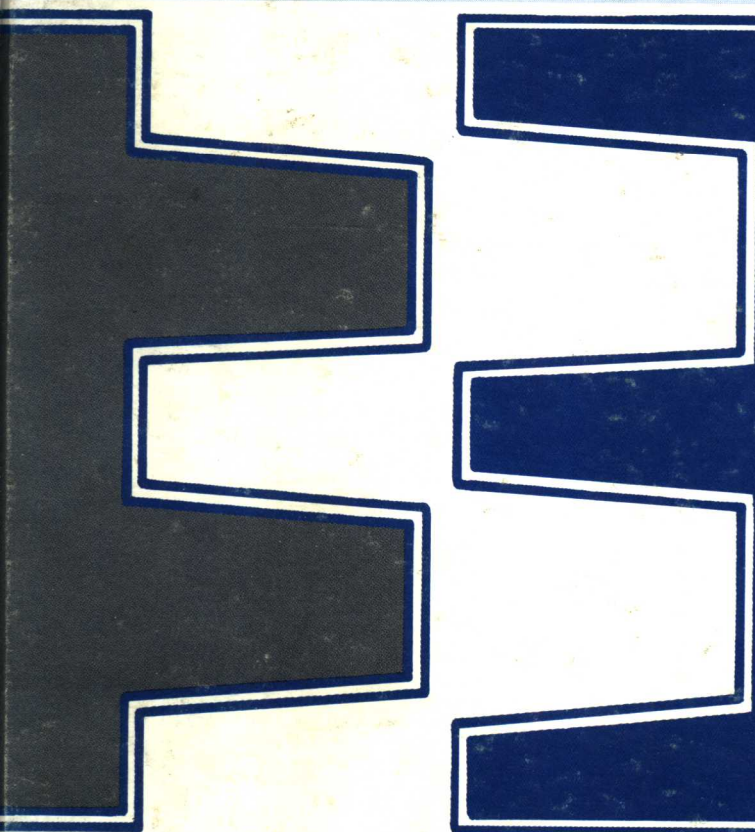


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HSLA STEELS **TECHNOLOGY &** **APPLICATIONS**



Conference Organized by
Michael Korchynsky

CONFERENCE **PROCEEDINGS**

AMERICAN SOCIETY FOR METALS

HSLA STEELS TECHNOLOGY & APPLICATIONS

Conference organized by:

Michael Korchynsky

Conference Proceedings of
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FOREWARD

This Proceedings of the International Conference on Technology and Applications of High Strength Low Alloy Steels, held October 3-6, 1983 in Philadelphia, Pennsylvania, contains close to one-hundred technical papers by metallurgists, scientists, and engineers representing four continents. The broad international interest in this conference attests to the growing acceptance of HSLA steels as a new class of engineering materials. These steels are no longer merely novel, but -- because of their cost effectiveness -- have become essential alternatives.

Of the many symposia and meetings devoted during the past decade to HSLA steels, "Microalloying 75" (October 1975, Washington, D.C.) is frequently referred to as a true milestone. In the postscript to Proceedings Microalloying 75, this thought was expressed: "Considering the over-all potential of microalloyed steels, the current 'state-of-the-art' represented by this volume may well be just the tip of an iceberg."

Developments during the intervening eight years seem to support this prediction. A number of these developments were covered at this Conference. A few randomly selected examples:

- ° Application of ladle metallurgy and injection technology to microalloying steels.
- ° Controlled (accelerated) cooling of plates and bars.
- ° Means for preventing occurrence of mixed grain sizes.
- ° Technology for production of coiled plates.
- ° Recrystallization Controlled Rolling.
- ° High Strength and dual phase continuously annealed sheets.
- ° Utilization of titanium nitrides and vanadium nitrides.
- ° Reduction in the carbon content of high strength steels, exemplified by ultralow carbon bainite steels (ULCB).
- ° Microalloyed bars and long products.
- ° Precipitation hardenable forging steels.

The foregoing examples illustrate the versatility possible in the design, production, and processing of microalloyed steels. The resulting products can be tailored to meet the specific needs of a given application in a most efficient way.

It is hoped that this volume will contribute to a more universal acceptance of HSLA steels by users, and -- like Proceedings Microalloying 75 -- will provide a stimulus to further perfection and expansion of this class of engineering materials.

As general chairman of the 1983 HSLA Conference, I would like to acknowledge the outstanding contributions of the Organizing Committee: J. D. Boyd (Canada Centre for Mineral & Energy Technology), A. T. Davenport (Republic Steel Corporation), A. J. DeArdo (University of Pittsburgh), G. M. Faulring (Union Carbide Corporation), F. B. Fletcher (Climax Molybdenum Company), S. S. Hansen (Bethlehem Steel Corporation), H. I. McHenry (National Bureau of Standards), P. E. Repas (United States Steel Corporation), G. H. Robinson (General Motors Corporation), and H. Stuart (Niobium Products Company).

And, in turn, the Organizing Committee must acknowledge the invaluable assistance provided by Barbara L. Thomas, who worked long and hard as coordinator between the committee and conference paper authors.

Credit for guidance and assistance is also gratefully extended to ASM management -- specifically, Edward L. Langer, Allan Ray Putnam, and Robert C. Uhl -- and to other Technical Division's staff -- Margaret Ternovacz, Lana Loar, and Shari Gerstenberger.

Michael Korchynsky
Union Carbide Corporation
General Chairman, 1983 HSLA Conference
May 1984

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THE SPECTRUM OF MICROALLOYED HIGH STRENGTH LOW ALLOY STEELS

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A review, by no means exhaustive, has been presented of the evolution of the use of micro-alloying additions over recent years to many different types of steels apart from the conventional HSLA steels. Some of the salient features of advances in the use of the current metallurgy of the ferrite-pearlite steels, and in particular of modern thermo-mechanical processing have been summarised, and the many roles of micro-alloying additions have been identified. Specific examples of line-pipe and hot rolled strip have been given, and the evolution of the microstructure has been described. The importance of the transformation characteristics of the austenite has been emphasised. The quenched and tempered and cold worked and annealed HSLA steels have been discussed. Attention has then been turned to the metallurgy of the acicular ferrite steels and of the currently important dual phase steels, in which recent metallurgical understanding has been discussed. The paper then goes on to discuss HSLA steels used for rod and bar products, and the recently developed micro-alloyed pearlitic forging and rail steels. The paper concludes with a consideration of the use of micro-alloying for enhanced hardenability and temperability in quenched and tempered engineering steels. Finally some aspects of modern steelmaking techniques and their implications are mentioned.

THE LAST 20 YEARS has seen an unsurpassed revolution in steel developments. This has been concerned less with the development of new compositions such as maraging or TRIP steels but rather with the application of existing knowledge to improve the response of materials to processing and to incorporate into existing materials the fruits of research into the underlying metallurgical principles. A major field in which this has occurred is High Strength Low Alloy Steels, and the importance of these materials to a wide range of industries can be judged by the plethora of conferences over the

past few years (1)-(28). The pace of development, and the volume of published work is bewilderingly great, so much so that it becomes virtually impossible to incorporate and assimilate all the new information into a rational framework for discussion. This is especially the case for high strength low alloy steels which have diversified from what were essentially carbon-manganese steels micro-alloyed with Nb, V or Ti into acicular ferrite or low carbon bainitic steels, quenched and tempered steels, cold rolled and annealed steels, dual phase steels, quenched and tempered steels, and the higher carbon more pearlitic steels, all containing one or more of the micro-alloying elements. In addition there have been major advances in the steelmaking of these materials, aimed at producing lower carbon contents without increased volume fractions of oxide inclusions, lower sulphur contents and the control of inclusion shape by yet a new generation of micro-alloying additions such as Ca, Zr, and rare earths.

The purpose of this paper can be no more than to introduce the general theme of the conference by a highly selective and by no means comprehensive account of some of the more relevant metallurgical phenomena involved in the present spectrum of high strength low alloy steels, possibly drawing attention to potential new developments and indicating some of the more important applications. To provide a metallurgical framework the various types of microstructure will generally be considered, as the steels used for different product forms or for varying applications may comprise one of several of the microstructural types. In some cases however it is necessary to consider generic types of steels. But all the steels of the high strength low alloy categories which will be discussed have several common aims, namely to improve the strength-toughness-formability-weldability combination of properties as economically as possible by utilising energy effective mill processing without the need for additional heat

treatment procedures, and to conserve ever-increasingly scarce or strategically important alloying elements by the use of micro-alloying additions and appropriate steelmaking additives. Not all these objectives can be met in each category of steel.

CONVENTIONAL MICROALLOYED HSLA STEELS

FERRITE-PEARLITE STEELS - The structure - property relationships in these steels are well established (29)-(32) as summarised in fig.1, and only minor refinements may be expected. Already the yield stress can be predicted to within limits governed by the accuracy of the testing techniques and the homogeneity of commercial steel. One feature not fully quantified is the effects of Fe_3C particle size at the ferrite grain boundaries. These carbides can become larger during the slow cooling of thick plate or during tempering for weld stress relief. It is well known that the ductile-brittle transition temperature (DBTT) is increased as these carbides become larger (33), fig.2, but recent work also shows that they influence the yield and tensile strength (34). The reason for this latter effect is as yet uncertain as also are the compositional and heat treatment conditions which can eliminate such large carbides, but increasing manganese is certainly an advantage (35).

Controlled Processing - The major question concerns the evolution of the polygonal ferrite structure, particularly during thermo-mechanical processing. The essence of controlled rolling and controlled cooling is that the structure of the austenite is conditioned to allow it to transform to ferrite with the finest possible grain size, in order to produce the greatest strength and optimum toughness and ductility. At the same time, the microalloy carbide/nitride should precipitate during or after the transformation to further enhance the strength, albeit with the sacrifice of some toughness. What is therefore required is a high ferrite nucleation rate and a low growth rate, and indeed the nucleation and growth of the microalloy carbides/nitrides are also of the utmost importance. In fact, nucleation and growth phenomena are central to thermo-mechanical processing, as they are also used to control recrystallisation during hot working (36)-(38).

The requirements to produce the necessary fine ferrite grain size are a fine austenite grain size, or 'pancaked' unrecrystallised grains, as these provide the greatest area of austenite grain boundary for ferrite nucleation, which can also occur on deformation bands in unrecrystallised austenite, on recovered substructure boundaries particularly if these contain precipitates, and on undissolved carbide/nitride particles (39). Consequently the two conditions to be aimed for are either the finest possible recrystallised austenite or heavily deformed but unrecrystallised austenite. It has long been established that the rate of

recrystallisation, and the grain size of the recrystallised austenite during hot rolling follow the classical laws of recrystallisation behaviour. Consequently a small starting grain size would be beneficial, fig.3, and one of the effects of microalloying additions is to achieve this by particles of carbides/nitrides restricting austenite grain growth at the reheating temperature. The stability, and solubility in austenite of the various microalloying carbides and nitrides are, however, very different, fig.4. Nitrides are more stable than carbides, and the stability increases, or the solubility decreases, in the order vanadium, niobium, titanium. Consequently, whereas the earlier generation of HSLA steels usually contained only a single microalloying addition, current steels tend to employ combinations, eg Nb-V. The NbC or VN (using enhanced nitrogen) tend to restrict grain growth, whilst the more soluble VC is used to precipitation strengthen the ferrite. More recently attention has been turned to the potential use of Ti in combination with V in order to take advantage of the greater stability of TiN and its slower growth rate, which provide it with the capability to act as a grain growth inhibitor at high reheating temperatures prior to hot working. The amounts of Ti needed are small, 0.01-0.015%, and whilst control of such additions is difficult, it is by no means impossible using currently available ladle injection techniques. Too large a titanium addition must be avoided otherwise large TiN particles are formed which do not restrict austenite grain growth. The extent of inhibition of grain growth is critically dependent on the dispersion of the TiN, and indeed of any other grain growth inhibiting precipitate, variations in the size distribution and volume fraction being able to change the grain coarsening temperature by up to 300°C for the same steel composition. Consequently attention must be paid to the effects of casting conditions on the precipitate distribution in continuously cast products, and to the effect of hot working prior to the slab stage in conventional ingot products.

Microalloying additions also have another important effect during controlled rolling, in that they retard recrystallisation. Much work has been done to identify the nucleation sites for recrystallisation (37)-(39), and it is now recognised that whilst the deformed austenite grain boundaries are the preferred nuclei, recrystallisation can also nucleate at the interfaces between recrystallised and unrecrystallised grains, on deformation bands in deformed coarse grains and on large microalloy carbide/nitride particles $\sim 0.5 \mu\text{m}$ in diameter. The rate of recrystallisation in C-Mn steel is very rapid, but can be markedly retarded by the microalloying additions, fig.5, and also by Mo and Mn. There have been extensive studies of austenite recrystallisation and on the precipitation effects of microalloy carbides/nitrides in deformed austenite (40)-(45). The long

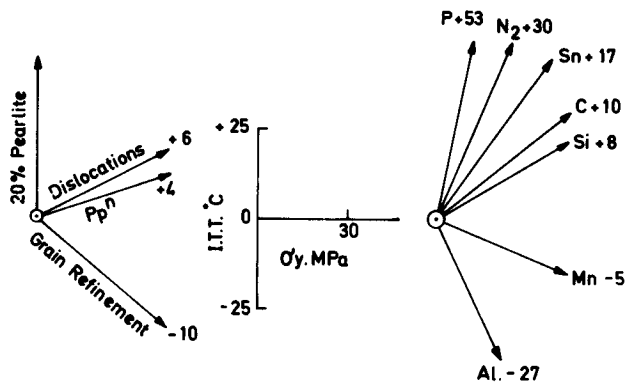


Fig. 1 - Strengthening mechanisms in polygonal ferrite and their effect on the ductile-brittle transition temperature. (Ratios indicate change in DBTT for 15 MPa increase in yield stress)

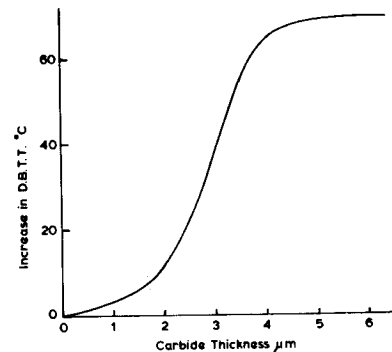


Fig. 2 - Effect of carbide size on the increase in ductile-brittle transition temperature in polygonal ferrite structures.

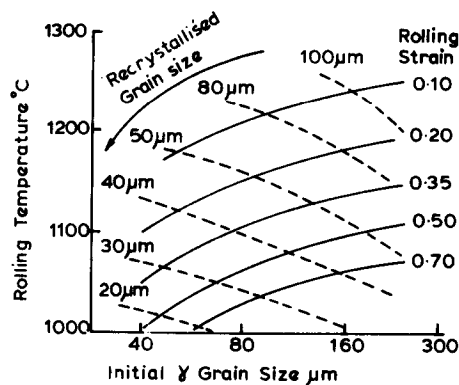


Fig. 3 - Effect of initial austenite grain size, rolling temperature and rolling strain on the recrystallised austenite grain size in Nb treated HSLA steel.

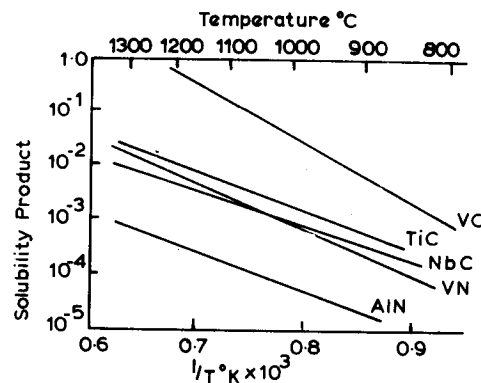


Fig. 4 - Solubility relationships for micro-alloy carbides and nitrides.

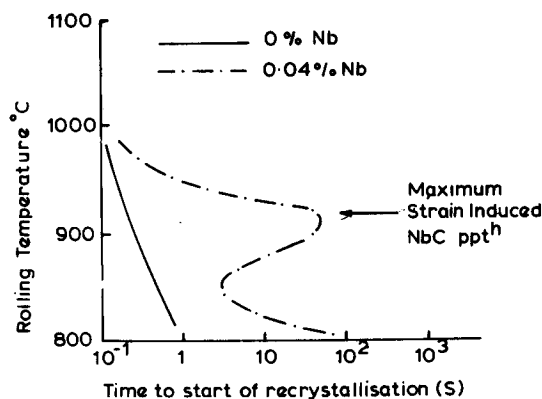


Fig. 5 - Effect of Nb on the retardation of the recrystallisation of austenite during hot working.

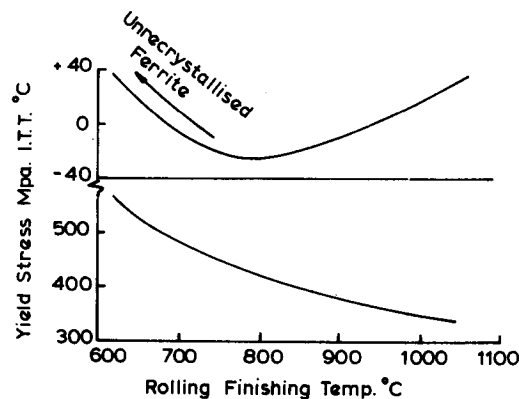


Fig. 6 - Effect of rolling finishing temperature on the yield strength and ductile-brittle transition temperature in the Nb treated HSLA steel.

standing controversy as to whether this retardation of recrystallisation is the result of solute or precipitate now seems to have been resolved. There is an accumulated body of evidence to show that dissolved solute atoms can retard both recovery and recrystallisation (39)(42) but that a major effect is produced by precipitation in the unrecrystallised austenite (39)-(45). This strain induced precipitation in austenite pins recovered sub-boundaries and inhibits recrystallisation. In order to do this, strain induced precipitation must occur prior to any substantial recrystallisation. Complications can be introduced by the different effects of strain and strain rate on precipitation and recrystallisation kinetics and particularly by the different solubilities of the various microalloy carbides and nitrides. It has been shown that the retarding effect of dissolved solute atoms on recrystallisation increases with increasing strain introduced by the solute into the austenite lattice, which influences the dislocation-solute interaction (46)(47). The effect seems to be, in ascending order of effectiveness Mn, Al, V, Mo, Nb and Ti, on an atomic basis. The effect of microalloy carbide/nitride precipitation on the retardation of austenite recrystallisation is less easy to predict because it depends on the intensity of the precipitation which in turn depends on the strain applied and on the composition of the steel. Also, the overall retardation is compounded of both the solute and precipitation effects. The intensity of the precipitation effect depends on the solubility of the microalloy carbide/nitride, the greater the solubility the less the intensity of strain induced precipitation. Hence VC is less able to retard recrystallisation than VN at the same temperature. Whilst VN and NbC have reasonably similar solubilities, niobium seems to be much more effective in retarding recrystallisation than does vanadium because of the much greater solute effect of Nb (44). But the intensity of precipitation also depends on the ratio of the microalloying addition to either carbon or nitrogen with respect to the stoichiometric ratio of the appropriate carbide/nitride phase. Because the temperature dependence of the solubility is maximised at the stoichiometric ratio (48), the nearer the steel composition is to that ratio the greater will be the intensity of strain induced precipitation and the more retardation of recrystallisation would be observed. As all HSLA steels are hypostoichiometric; a decrease in carbon content would therefore be expected to produce more strain induced precipitation, and hence greater retardation of recrystallisation. The overall alloy content of the steel also is important, carbide or nitride forming elements such as Mn and Mo decrease the activity coefficient of C or N but increase that of Nb and V (41)(44)(45)(49-52). Because the effect on the activity of the interstitial element predominates, Mn and Mo increase the solubility of both VN and NbC (53)-(55), thus retarding their precipitation by

lowering the supersaturation and giving rise to less strain induced precipitation. Hence both Mn and Mo induce less retardation of recrystallisation by strain induced precipitation and V and Nb can react similarly (41).

The importance of the retardation of recrystallisation during controlled rolling lies in the ability to use a low finishing temperature to produce elongated unrecrystallised austenite grains which can transform to very fine polygonal ferrite. This is one of the variants of controlled rolling which is derived from the classical work describing the main regimes for hot working austenite (37)(38). However, the different microalloying elements lead to different finishing regimes to produce the heavily elongated unrecrystallised grains because the temperatures at which the different carbides/nitrides are predominantly strain induced to precipitate varies; for example, most rapid strain induced precipitation occurs for TiC/TiN at 1025°C (56), for NbC (41)(44)(45)(57) and VN at 900°C (44)(58) and for VC at 850°C (41)(44)(45)(58).

Finishing temperatures below the A_{r3} are sometimes used and can give a marked increase in yield strength. The ferrite formed during rolling is deformed but does not recrystallise because of the inhibiting effects of microalloy carbide/nitride. This ferrite is therefore only recovered, but the ferrite formed from the transformation of the unrecrystallised austenite is the normal dislocation-free polygonal ferrite. The higher strength is said to be due to the fine sub-grain structure in the deformed but unrecrystallised ferrite, and the very fine polygonal ferrite formed during the transformation.

Variable effects on the DBTT have been reported. In some cases the DBTT is lowered, possibly due to textural effects producing splitting or separations, and the heavily polygonised sub-grain structure in which the sub-grain boundaries are relatively high angle and thus behave in effect as grain boundaries (59). On the other hand, such a low finishing temperature has been reported to increase the DBTT, fig.6, probably due to insufficient deformation resulting in less pronounced texturing and much lower angle polygonised sub-grain boundaries, (29)(35)(60) or due to the deformed ferrite not being recrystallised and giving high strength. It seems that this type of controlled rolling could be very difficult to control, just as is continuum rolling (61) in which very heavy deformations at temperatures as low as 400-500°C in the ferrite region are used either to produce a heavily textured, very fine recrystallised ferrite or a fine polygonised, high angle, recovered ferrite sub-grain structure. All these effects increase the strength and decrease the DBTT.

The solubility, stability and precipitation kinetics of all the microalloy carbides and nitrides are different. The nitrides are much more stable and less soluble than the carbides

(62). Nitrogen, therefore, which has long been used in V steels, may become increasingly important in all microalloyed HSLA steels, especially if advantage is to be taken of the high stability and slow coarsening rate of TiN. The solubility of microalloy nitrides is about two orders of magnitude less than that of their carbides in both austenite and ferrite (62). Also they are at least an order of magnitude less soluble in ferrite than in austenite (58) (62) and so the supersaturation and driving force for precipitation is greater for nitrides. Thus nitrides can occur in finer dispersions (63) (64) which may, if strain induced in austenite, give more retardation of recrystallisation and inhibition of the growth of recrystallised grains. Because they also occur in larger volume fractions due to their lower solubility, they may if formed in the ferrite lead to greater precipitation strengthening. In fact nitride dispersions in creep resisting alloys have been found to be very effective, and it may well be that greater attention to precipitation of nitrides in both austenite and ferrite would be rewarding.

The current use of multiple microalloying additions has led to the need to identify more precisely the precipitating phases. The belief that the carbides and nitrides of a given microalloying addition were completely mutually soluble to give a carbo-nitride of composition reflecting the carbon and nitrogen contents of the steel, must now be questioned. It has been shown for example that there are firm thermodynamic reasons (62)(65) for suspecting that nitrides precipitate first, and when the nitrogen is used up, carbides precipitate. With multiple microalloying additions the situation becomes more complex, especially as under completely equilibrium conditions, many of the microalloyed carbides and nitrides are all mutually soluble to some considerable extent (66) and indeed such mutual solubility has been observed using modern high resolution electron-optical analytical techniques (67). Various models have been proposed in an attempt to predict precipitation sequences (67)(68), using assumptions of either complete immiscibility, complete miscibility or co-precipitation effects. These models tend to employ 'equilibrium' solubility data, and do not consider the kinetic effects which are so important. They are worthy of much more study. In particular, the suggestions that various of the microalloy carbide/nitrides contain iron, and that the iron content increases as the particle size decreases (69)(70) requires careful investigation, as if true it could call into question the validity of much of the work on solubility studies, to say nothing of the relationship between the response of controlled rolling to stoichiometry. Such a study is even more important if it is supposed that future generations of HSLA steels will use even more complex combinations of microalloying additions, each of which has its own 'C' curve

for carbide and/or nitride precipitation in both austenite and in ferrite. Such precipitation 'C' curves, each occurring in particular temperature ranges, may then allow the optimum effects of each microalloy addition to be employed to the full in each controlled rolling regime.

Two points require to be emphasised:

- (i) strain induced precipitation in the austenite decreases the amount of precipitate available to form during and after the transformation to ferrite, and so detracts from the precipitation strengthening. However, besides preventing grain growth of the recrystallised austenite, strain induced precipitates and microalloy elements in solution both increase the yield stress of the austenite (43)-(45). When the finishing temperature is low, this can materially increase the loads on the rolling mill.
- (ii) undissolved microalloy carbide/nitride precipitates at the reheating temperature contribute nothing to strength and are generally so large as not to cause grain boundary pinning unless they are in very large volume fractions. This can be seen from the predictions of the Gladman model (71) for grain growth inhibition, fig.7. Consequently, with the exception of TiN which is very stable and grows very slowly there is no great benefit substantially to exceed the solubility at the reheating temperature. But the nearer the steel composition is to stoichiometry the more strain induced precipitate will be formed to inhibit recrystallisation and grain growth of recrystallised grains, and also the greater will be the potential for precipitation strengthening of the ferrite. Undissolved carbides/nitrides $>0.5 \mu\text{m}$ in diameter, are however capable of accelerating recrystallisation by particle stimulated nucleation (39). As will be discussed later, in some cases this could be useful.

Controlled Rolling Methods - A recent review of controlled rolling (59) has identified the methods which can be used to condition the austenite to produce the finest ferrite grain size and optimum precipitation strengthening. These are:

- (a) a low reheating temperature to produce a fine initial austenite grain size, but this can decrease the potential for subsequent precipitation strengthening.
- (b) Austenite grain refinement by recrystallisation which involves:
- (c) Suitable pass schedules and reductions to obtain in the initial passes a fine, uniform recrystallised austenite.
- (d) Delay between roughing in the recrystallisation regime and finishing in the unrecrystallised regime.
- (e) Suitable reductions in the unrecrystallised regime and in some cases finishing

below Ar_3 , although this can be fraught with difficulties.

As there are many combinations of these factors which can be successfully employed, each mill has its own preferred controlled rolling route. However it is useful to identify two fairly typical processes applicable to different products.

(a) Line pipe is usually made on a plate mill, the difficulty being to give sufficient reduction per pass in the initial passes. The time between passes can produce interpass recrystallisation and uneven grain growth, so that a mixed grain structure is developed which can lead to inferior toughness. The steel is then allowed to cool into the unrecrystallised regime, where large deformations are often emphasised to produce very thin unrecrystallised grains which mitigate to some extent the problem of a mixed austenite grain size. This type of process slows production and is more difficult where size changes are frequent, but is economic on custom designed mills and also, incongruously, on old mills which have a sufficiently slow production rate not to be inconvenienced by controlled rolling delays (35). As plate thicknesses increase it becomes increasingly difficult to produce the requisite deformations, to accept the holding delays and to withstand the increased mill loads.

(b) Hot rolled strip for formable applications employs a different process route because it is possible to employ accelerated cooling on the run-out table, followed by slow cooling in the coil. Roughing and finishing can employ very heavy reductions and the interpass time is very short, so that there is little time for recrystallisation and grain growth between passes. In the roughing train dynamic recrystallisation is common, whilst in the finishing train the deformation is more likely to be in the unrecrystallised regime. It is possible to force recrystallisation to a very fine austenite grain size, even in the finishing train, by the use of large reductions.

Because the effective use of controlled rolling either to produce the finest recrystallised austenite or thin elongated unrecrystallised grains, prior to transformation, requires heavy reductions late in the rolling process, work is now being carried out to investigate how very fine recrystallised austenite grains may be produced without resort to heavy deformations at low temperatures. In the past this was attempted by introducing large volume fractions of undissolved Nb(CN) by increased niobium contents, and low reheating temperatures with obvious loss of precipitation strengthening. Also, large volume fractions of NbC could lead to hot tearing problems during continuous casting. An alternative approach is to use a fine dispersion of a very stable precipitate, such as TiN, which is slow to coarsen and which will

allow high reheating temperatures to be used to dissolve vanadium or niobium carbides/nitrides for adequate precipitation strengthening, and yet preserve a fine austenite grain size prior to rolling. By rolling, albeit with high deformations at high temperatures, such a fine initial grain size will produce very fine recrystallised grains which also will be stabilised against grain growth by the TiN or by an appropriately strain induced precipitate. This so called 'recrystallisation rolling' (72) may enable many of the economic and production difficulties of current controlled rolling practices to be overcome.

The Evolution of the Austenite Micro-structure - During controlled rolling the paramount requirement is the production of conditioned austenite to produce the finest possible ferrite grain size. Attempts are therefore being made to predict and calculate the evolution of the austenite grain size and morphology during the controlled rolling process (73)-(78) and also the ferrite grain size developed from the austenite during transformation (76)(77). These studies use the temperature and stress dependence of the strain rate, and strain during a pass on the recrystallisation rate and recrystallised grain size, the effect of initial grain size on recrystallisation kinetics and recrystallised grain size, and the kinetics of grain growth during the interpass interval, to calculate the grain size developed during each pass of a rolling sequence. Equations are produced describing the dependence of the recrystallised grain size on the initial grain size, the strain or strain rate, and the temperature. Some of these equations are produced by empirical analysis of data, whilst others are developed from basic principles. The equations can then be fed into a computer program for calculating the recrystallised grain size after each pass and hence the evolution of the austenite structure.

A further refinement is to allow for the elongation of unrecrystallised grains in terms of an 'effective' grain size which incorporates the effective austenite grain boundary area. Consideration of rolling in the non-recrystallisation regime is possible (38)(79). The evolved austenite grain size can then be used to predict the ferrite grain size, as will be discussed later.

In order to apply these sophisticated calculations to an actual rolling mill requires that the strain, strain rate and temperature in each pass are not only known but are also constant. Simply to obtain data on these parameters for a commercial mill is not easy, and it has also been shown recently that major variations in the parameters can occur transiently during any particular rolling pass (80). Clearly much more work requires to be done in this important field of study, and it is not beyond the bounds of possibility that by making instantaneous measurements on a mill during actual rolling it may be possible by a com-

puterised feedback technique to continually adjust the rolling parameters not only simply to control the physical dimensions and shape of the rolled product, but also to continually monitor and control the evolution of the microstructure.

Transformation of the Austenite - The requirement for the finest ferrite grain size necessitates a high ferrite nucleation rate during the transformation. As ferrite nucleates predominantly at austenite grain boundaries, the need for a fine austenite grain size is apparent. The parameter which has been often used to indicate the nucleation frequency for ferrite is the austenite grain boundary area per unit volume, S_V , (38) which increases with a refinement of the recrystallised austenite grain size and with increasing elongation of unrecrystallised austenite grains, fig.8. It has been suggested that because the driving force for ferrite formation is small, a few hundred $J\ mol^{-1}$ even at large undercooling, only the most potent nucleation sites will be effective (81)(82). As it was shown that for a given S_V , the ferrite grain size was finer when produced from unrecrystallised austenite than from recrystallised austenite (38)(79), fig.9, then it seemed that ferrite nucleated more potently at the deformed austenite boundaries. This was attributed to the serrations or bulges at such boundaries (81) acting as particularly potent nucleation sites, and so it was concluded that elongated austenite grains were highly desirable. This is in fact so, particularly if they are so thin that the ferrite grains from opposite boundaries can impinge with minimum growth. However more recent work (77)(79) has shown that at very high S_V values, i.e. small austenite grain sizes, the ferrite grain size depends only on S_V irrespective of the austenite grain shape. Thus for the most effective ferrite refinement, deformation below the austenite recrystallisation temperature may not be necessary; all that is required is a very fine recrystallised austenite grain size. Hence the potential for 'recrystallisation rolling'. The various nucleation sites for ferrite have now been well documented (79). They are not only austenite boundaries, but also deformation bands, second phase particles (particularly undissolved micro-alloy carbides/nitrides), recovered sub-grain boundaries especially if decorated by precipitates, and even the non-coherent ledges on twin boundaries. A mechanism is now also available to explain multiple or cascade ferrite nucleation (79) which has long been known to be associated with deformed austenite grain boundaries. Because many of these sites occur particularly in unrecrystallised austenite, the finest ferrite grain size will be produced from the finest austenite grain size rolled to a maximum extent below the recrystallisation temperature (38), fig.10. An important question however is how fine must the austenite be, which depends on the ratio of austenite grain size to ferrite grain size. As shown in fig.9, the ferrite grain size can almost equal the austenite

grain size if the latter is very fine, but for large austenite grain sizes the ferrite grain size may be less than one tenth of that of the austenite (79).

A useful method of refining the ferrite grain size for a given austenite grain size is to decrease the transformation temperature. This increases the ferrite nucleation rate and the effect may be achieved by alloying (the reason for the high manganese content in HSLA steels) or by increasing the cooling rate. Care must be taken not to depress the transformation temperature too far, otherwise bainite may form and lower the yield stress, as well as impairing toughness. However, the introduction of bainite into the structure also eliminates discontinuous yielding and minimises the Bauschinger effect, which has advantages in line-pipe production (35). Bainite is more readily produced in steels which contain small molybdenum additions which suppress pearlite formation and result in structures giving polygonal ferrite with bainite or acicular ferrite (35).

An interesting feature is the increase in the transformation temperature which occurs when unrecrystallised austenite is transformed at a given cooling rate. This can be explained by the acceleration of the transformation by strain inducement. It might be expected that this increase in transformation temperature would produce coarser ferrite, but this is not the case, presumably due to the much greater overall ferrite nucleation rate in unrecrystallised austenite. Studies of the rates of nucleation and growth in austenite deformed and recovered to various extents would be useful.

In strip rolling, the transformation temperature is to a large extent controlled by the water cooling on the run-out table which is used to determine the coiling temperature. The greater the cooling rate or the lower the coiling temperature, the finer is the ferrite grain size formed from a given austenite structure. Fig.11 shows how the ferrite grain size depends on the transformation temperature, but such a relationship will vary for different steel compositions and austenite grain sizes. The coiling temperature is also matched to the finishing rolling temperature, to produce the required ferrite grain size, a typical diagram showing this in fig.12 (83).

In plate rolling, the cooling rate is largely controlled by the plate thickness. Due to practical difficulties in controlled rolling the thicker plates, these are often normalised. In order to optimise the strength and toughness some form of accelerated cooling is not infrequently used to increase the cooling rate, decrease the transformation temperature and refine the ferrite grain size. As will be shown later, the transformation temperature also controls the size and distribution of precipitates in the ferrite. Recently an equation has been obtained which defines the ferrite grain size in terms of the recrystallised austenite grain size and the rate of cooling (76). In