The Formation of Panel Cracks in Steel Ingots: A State-of-the-Art Review

I. Hot Ductility of Steel

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ABSTRACT

To provide a fundamental understanding of panel-crack formation in ingots, the hot ductility of steel is reviewed in the first of a two-part paper. Three zones of reduced ductility can be identified at elevated temperature; two of these, in "low" and "intermediate" temperature ranges, contribute to the formation of panel cracks. The low-temperature zone occurs in the two-phase austenite-to-ferrite region below the Ar_3 temperature. It results from strain concentration in the films of primary ferrite forming at austenite grain boundaries. The primary ferrite encourages preferential precipitation of nitrides which exacerbates the strain concentration and grain-boundary embrittlement. The intermediate-temperature zone of reduced ductility extends from the Ar_3 temperature to as high as 1200°C. Phases, principally nitrides, precipitating at austenite grain boundaries play a major role in the ductility loss. Creep-type fracture occurs due to coalescence of cavities nucleating at the grain-boundary precipitates. Thus the presence of strong nitride formers such as Al, Nb and B in excess of critical concentrations markedly reduces the ductility. The third zone of low ductility is found at temperatures within 30 to 70°C of the solidus and is due to the presence of interdendritic liquid films rich in S and P.

INTRODUCTION

Despite the advantages of the continuous-casting process, over two-thirds of world steel production currently follows the conventional ingot casting route. Although the adoption of continuous casting is accelerating, static ingot casting will continue to be an important mode of steel production for decades to come.

The quality of ingots is a matter of great concern, particularly so because defects can deleteriously affect the yield of the energy intensive casting process. One serious quality problem that has been nagging the steel industry in at least seven countries for over 40 years is the formation of panel cracks. The term "panel" describes the location of the cracks which frequently appear in the concave, panel areas on fluted or corrugated ingots. However, this defect has also been called center face cracking,[1] longitudinal cracking,[2-8] pearlitic cracking,[4] cooling cracking, longitudinal surface cracking,[9] thermal stress cracking,[10] reheating cracking, phase transformation cracking, hair line cracking,[6-7] tortoise shell cracking,[11] transverse cracking,[7] ovally arranged cracking,[12] vertical cracking,[13] and even "crazy" cracking.[1-14] These different names give an indication of the many different manifestations of panel cracks and the variety of mechanisms that have been proposed to explain their formation.

Panel cracks appear in a variety of low- and medium-carbon killed steels, but are always associated with aluminum-treated grades and are greatly affected by the thermal treatment of the ingot. They have been found in a wide range of ingot sizes and shapes, from 1.5 ton square, flat billets to 30 ton rectangular, corrugated ingots. Round, fluted ingots also have been affected.[8]

The defect is characterized by one or more irregular, intergranular cracks which generally run longitudinally down the face of the ingot as shown in Figure 1. They extend to a considerable depth below the surface and travel along the austenite grain boundaries. The reasons for panel-crack formation are not fully understood and many complicated mechanisms have been proposed. However, it is generally agreed that the problem is caused by a combination of reduced intermediate temperature (600-900°C) ductility involving the presence of aluminum nitride, or "AlN," precipitates and stress generation due to both thermal contraction and phase transformation. Panel cracks usually

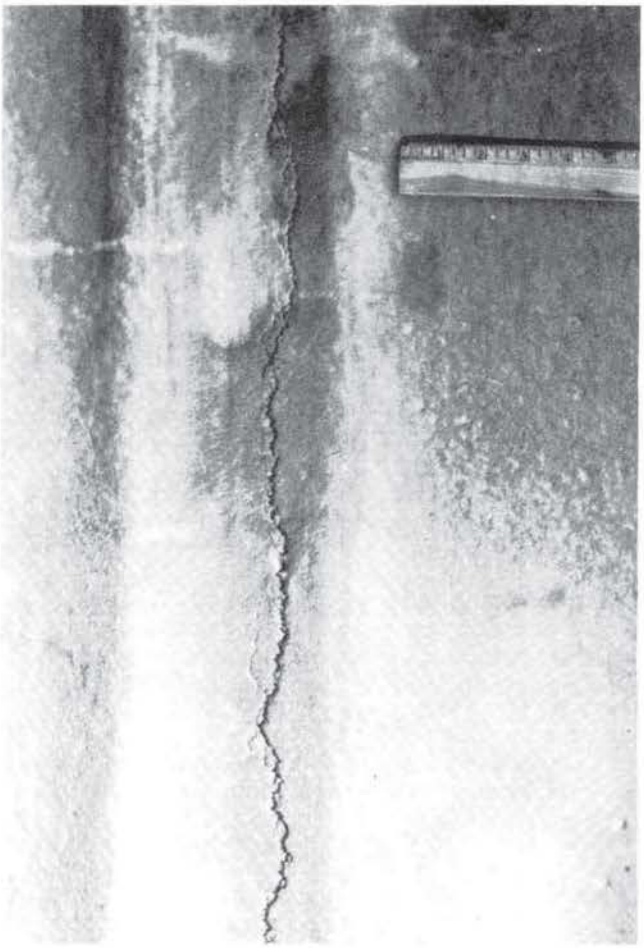

FIG. 1. Typical appearance of panel crack running along corrugations of a 760 × 1520 mm, 25 ton steel ingot.[14]

are not discovered until a much later stage in ingot processing, typically during rolling. They present a serious and expensive problem because affected slabs cannot be salvaged and must be scrapped; thus there is a strong incentive to discover methods to eliminate panel-crack formation.

The first part of the present work reviews the ductility of steel at elevated temperatures to provide a fundamental background with which to examine panel cracking itself. The second part then seeks to review knowledge of the occurrence and proposed mechanisms of panel cracking as well as solutions to the problem. This review is part of a larger project involving finite-element analysis, in which panel cracks are being related to the stress field generated in an ingot during the different processing stages.

OTHER CRACK PROBLEMS IN STEEL

At the outset it is important to distinguish between panel cracks and other types of cracks that form in ingots by different mechanisms. This is particularly important when so many studies on panel cracking refer to it by a different name. One of these different mechanisms is "hot tearing" or "hot shortness" which is responsible for transverse cracks in statically-cast ingots. It also gives rise to virtually all of the crack defects in continuously cast steel with the exception of transverse surface cracks. Cracks resulting from hot tearing are interdendritic and exhibit a smooth fracture surface, similar in appearance to panel cracks. They form during the early stages of ingot solidification in a zone of low ductility just below the solidus temperature. The stresses causing the cracks are usually generated by sticking or bending in the mold. Hot tearing is relatively insensitive to subsequent thermal treatment but is strongly influenced by the sulfur and phosphorus content and manganese/sulfur, or "Mn/S," ratio in the steel as well as conditions in the mold such as metal temperature, fill rate, mold design, and stirring.

Another type of cracking, often called "clinking" because it is audible, has a distinctly different mechanism from both panel cracking and hot tearing. Clinks appear only in high-carbon or alloy steel grades with high-carbon equivalents and are generated at lower temperatures (about 300 °C). Cooling clinks occur after an ingot has been stripped early and exposed to a cold atmosphere; and reheating clinks are formed when a cold ingot is charged into a hot pit and rapidly heated. Both types of clinks are thought to be caused solely by the generation of high thermal stresses in the outer ingot skin, and are not associated with either AlN or a ductility loss at intermediate temperatures.

A third type of cracking, "hydrogen flaking," also affects medium-carbon steel ingots and is closely associated with thermal treatment and the hydrogen content of the steel. Hydrogen flaking is caused by hydrogen gas nucleation in the solid steel, and can be controlled by lowering the hydrogen content or by holding at about 650 °C or slow cooling the ingot to facilitate hydrogen diffusion.

Turning from ingot casting to other steel treatment processes, many experience cracking problems with features similar to panel cracks. Several examples can be cited. Sand castings of carbon and low-alloy steels with high aluminum and nitrogen contents occasionally exhibit intergranular cracks known as "rock candy fracture."[15-17] Lorig and Elsea[15] concluded in 1947 that AlN precipitation at primary austenite grain boundaries was the principal cause of this fracture. Woodfine and Quarrel[16] subsequently confirmed this mechanism and added that the problem was most severe in large castings where a slow cooling rate took place after solidification or the temperature was held between 800 and 1100 °C. A similar mechanism accounts for "surface break up" in large, aluminum grain-refined, low-alloy, nickel-bearing, steel forgings.[18] Nickel causes the solution temperature of AlN to increase by about 100 °C which then allows the precipitation of AlN at forging temperatures. This reduces the hot workability of the steel and results in cracking.[18] "Y" crack formation during rolling of low-carbon, alloy steels is also associated with AlN precipitation at austenite grain boundaries. Chuen[19] determined that the steel composition, teeming temperature, rolling velocity, track time and reheating practice most influenced this type of crack. Low-carbon, silicon-manganese steels are subject to "temper-embrittlement" during heat treatment or reheating, if nitrides such as AlN are allowed to precipitate.[20-22] Inoue determined that nitrogen segregation to the austenite grain boundaries causes severe embrittlement if the steel is tempered at 500-600 °C and quenched.[21] Another example can be found in the welding of low-carbon steels since cracks can form in the heat-affected zone.[23-25] Crack formation is significantly affected by cooling rate,[24] the AlN content[24] and a loss in ductility around 600 °C.[25] Transverse cracking is experienced in continuously cast steel slabs[26-31] if stresses generated during straightening occur when the surface temperature of the strand is in the intermediate-temperature range of reduced ductility between 700 and 900 °C. Also, it is suspected that continuously cast blooms may be subject to panel cracking after exiting the caster in the same manner as statically-cast ingots.[8]

Each of these defects involves AlN embrittlement as well as intergranular cracking along prior austenite grain boundaries, and is greatly influenced by the thermal treatment. These facts strongly imply that a similar mechanism is operating in all of these cracking problems as well as in panel cracking, despite the major differences between the metallurgical processes involved. The one factor apparently in common to each of these problems is a loss in the hot ductility of steel; therefore, the next section reviews studies made in this area. Particular attention is given to AlN precipitation and thermal history effects in the intermediate-temperature range. Emphasis is placed on those aspects which are most closely related to the formation of panel cracks.

HOT DUCTILITY OF STEEL

Several different methods have been applied to determine the hot ductility of steel. Tests have been made with an Instron machine and induction furnace but this traditional method has problems associated with premature necking. Alternatively, researchers have employed torsion-testing machines to achieve higher strains before fracture. However, the accuracy of both these methods has been questioned owing to the difficulty of reproducing an "as cast" structure by reheating solid material from ambient temperature. To overcome this, many workers have used a Gleeble machine, in which a specimen can be melted and resolidified "in situ," slowly cooled, and then tested, possibly representing a better simulation of true casting conditions. However, a disadvantage of this technique is that only a very small amount of material is tested (about 1 cm^3) so local nonuniformities can play a large role. The local nature of the test also makes it impossible to record actual load and elongation so that mechanical behavior must be inferred solely from reduction-in-area measurements and analysis of the cooled fracture surface.

Notwithstanding these difficulties, numerous studies have been performed on the hot ductility of steel which has been found to correlate remarkably well with a variety of cracking problems. The following sections will describe the different temperature zones of lowered ductility for plain-carbon and low-alloy steels.

Zones of Reduced Ductility

In general, the ductility of steel at elevated temperatures is excellent. However, there are at least two distinct temperature regions in which its ductility drops markedly. The first of these appears at high temperatures within about 50°C of the solidus temperature. The ductility in this zone is virtually zero and as mentioned earlier, is responsible for hot tearing.

The second range of reduced ductility extends from as high as 1200°C[32,33] to as low as 600°C,[30] as shown in Figure 2. In this broad interval, reduction in area or "R.A." can have almost any value, ranging from a minimum of about 10 percent to almost 100 percent. While the total strain-to-fracture can occasionally approach zero, there is always some local deformation[34] which distinguishes low ductility cracking in this region from that in the high-temperature zone. This second "ductility trough" can be divided further into at least two overlapping temperature zones in which different embrittling mechanisms operate. One of these affects steel while it is entirely in the austenite phase, and lies in an intermediate-temperature range from the Ar_3 temperature (usually about 800°C) to as high as 1200°C. Because the start of the austenite-ferrite or "$\gamma \rightarrow \alpha$" phase transformation can be substantially delayed below equilibrium in micro-alloy steels, this temperature region can extend to below 700°C. While the mechanisms operating in this region are poorly understood, they are related to precipitate pinning effects, grain-boundary sliding and delayed recrystallization. The other zone is closely associated with the $\gamma \rightarrow \alpha$ phase transformation and lies in a lower temperature range below the Ar_3. Strain concentration at ferrite networks surrounded by austenite, or pearlite at lower temperatures, are responsible for reduced ductility in this zone. This mechanism may extend to below the Ar_1 temperature where ferrite networks surrounded by pearlite may constitute a final zone of lower ductility.

Although innumerable investigations have been made on the hot ductility of steel below 1200°C,[1-3,7,15-18,24,29-51] few researchers have found two distinct intermediate-temperature ducitlity troughs at the same time. Either one, the other, or a combination of the two usually is observed, depending on the location of the Ar_3 phase transformation temperature with respect to the observed ductility trough. Although a ductility trough has been directly linked to the Ar_3 temperature,[37,48,49] other studies report it extends to much higher temperatures,[43,44] while still others find a trough entirely above the Ar_3.[29,32,34] Because the $\gamma \rightarrow \alpha$ phase transformation is influenced by alloying elements, cooling rate, strain rate and precipitate action itself, relating the loss in ductility to the Ar_3 temperature is often complicated. Thus, the division of the second ductility trough into two further zones is more due to a difference in fracture mechanisms and steel phases present than to separable ranges of temperature over which the mechanisms operate.

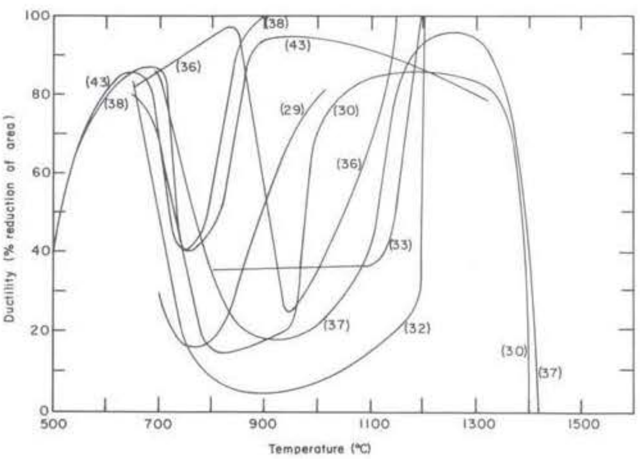

FIG. 2. Hot ductility of low-carbon steels containing manganese and aluminum as reported by various researchers.

These temperature zones of reduced ductility and their corresponding embrittling mechanisms are illustrated schematically in Figure 3. The next sections will elaborate on each in turn.

HIGH-TEMPERATURE ZONE

At temperatures just below the solidus, the strain-to-fracture of steel is less than 1 percent. Many studies have been conducted on this zone of reduced ductility,[47,52-62] and the mechanisms that are operative are probably the best understood. As depicted in Zone A of Figure 3, the ductility is reduced by the microsegregation of S and P residuals at solidifying dendrite interfaces which lowers the solidus temperature locally in the interdendritic regions. The ductility remains effectively zero until the interdendritic liquid films begin to freeze. Severe embrittlement is experienced at all temperatures above the "zero ductility temperature" which occurs within 30-70°C of the solidus as shown in Figure 4. Any strain that is applied to the steel in this temperature region will propagate cracks outward from the solidification front between dendrites. The resulting fracture surface exhibits a smooth, rounded appearance, characteristic of the presence of a liquid film at the time of failure.

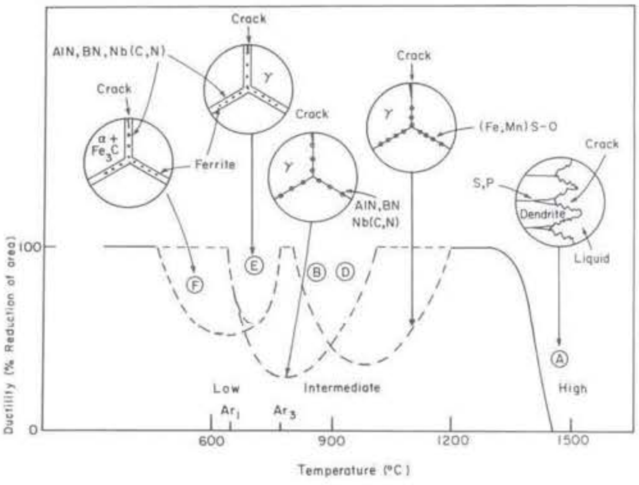

FIG. 3. Schematic representation of temperature zones of reduced hot ductility of steel related to embrittling mechanisms.

The transition from brittle to ductile behavior occurs over a narrow temperature range on heating.[52] However, on cooling, ductility may not approach 100 percent until somewhat lower temperatures are reached. In fact, Figure 4 indicates that some embrittlement may be encountered at temperatures as low as 1200°C. In theory, this embrittling mechanism can continue to operate to as low as the Fe-FeS eutectic temperature of 980°C.[32,34,63] Increased contents of S, P, Sn,[59] Cu[7,60] and Si[47] all worsen the ductility. Manganese is beneficial since it preferentially combines with S to form less harmful MnS precipitates, thereby preventing liquid film formation. A Mn/S ratio greater than 20 minimizes cracking by this mechanism,[57] partly by raising the ductile/brittle transition temperature.

This zone of reduced ductility is responsible for the problem of hot tearing as previously discussed. The ductility is relatively insensitive to subsequent thermal treatment[34] and strain rate.[52] In comparison, the fact that panel cracking is sensitive to the thermal history of the steel ingot suggests, in the absence of strain generation considerations, that the high-temperature zone of low ductility plays an insignificant role in the problem.

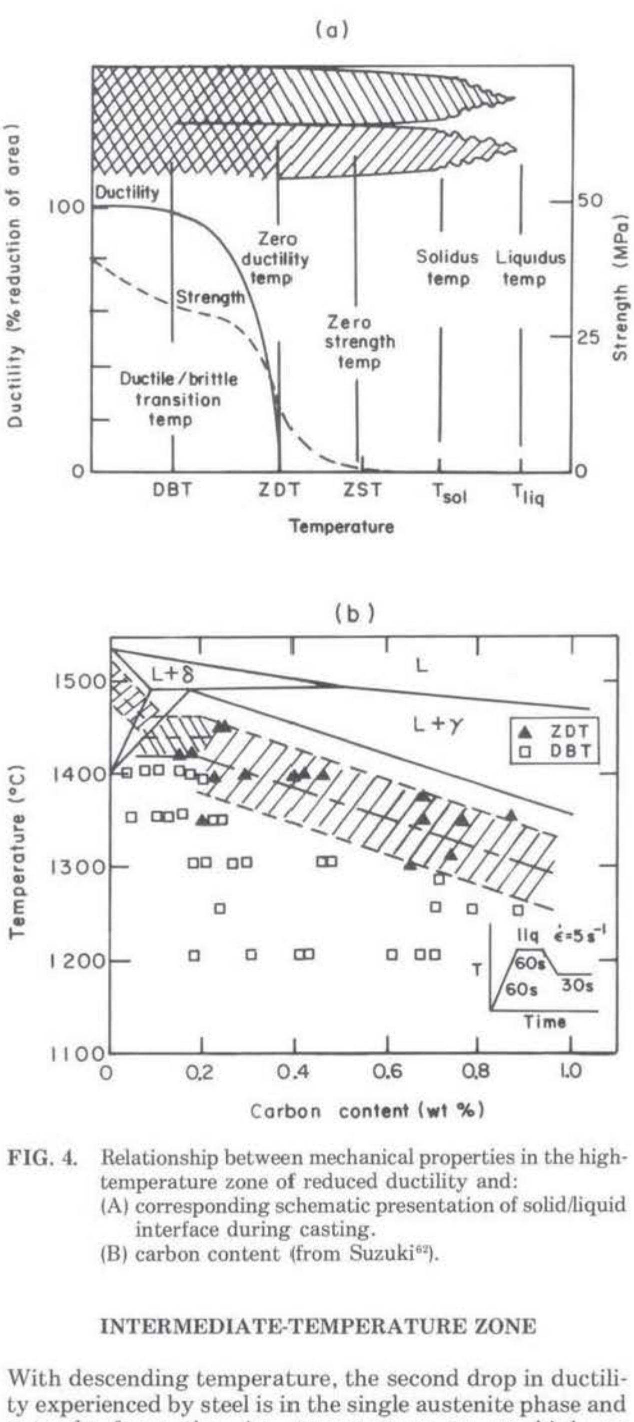

FIG. 4. Relationship between mechanical properties in the high-temperature zone of reduced ductility and:
(A) corresponding schematic presentation of solid/liquid interface during casting.
(B) carbon content (from Suzuki[62]).

INTERMEDIATE-TEMPERATURE ZONE

With descending temperature, the second drop in ductility experienced by steel is in the single austenite phase and extends from the Ar_3 temperature to as high as 1200 °C.[33,37] Above this temperature range, dynamic recrystallization occurs so readily that high ductility is assured, virtually unaffected by steel composition and processing conditions. While a great deal of study has been done on the ductility of austenite below 1200 °C,[18,24,29-50] elucidation of the mechanisms operating in this temperature region is incomplete owing to their complexity.

The ductility of steel specimens in this temperature range is directly reflected in the appearance of the fracture surface. High-ductility fractures are transgranular with characteristic dimples and a few large precipitates, indicating that fracture initiated at isolated inclusions dispersed throughout the matrix. In contrast, tests made in a temperature region of low ductility always exhibit an intergranular fracture along austenite grain boundaries making large angles with the major stress axis. In fact, the variation in RA values correlates well with the fraction of fracture surface occurring on austenite grain boundaries.[38]

The surface of specimens fractured in the intermediate-temperature zone of reduced ductility exhibits numerous precipitates of varying types including sulfides (Mn, Fe, and possibly Al[81]),[1,34,37,50] oxides (Mn, Fe and Al)[30,37] and nitrides (AlN[29,41-44] niobium carbonitride compounds, or "Nb(C,N)"[29,37,38,44,45,47,64,65] and BN[23,46]). Most of the fracture surfaces indicate a creep-type failure due to the coalescence of cavities nucleating at the grain-boundary precipitates.

Effect of Steel Composition

Steel composition is extremely important in determining the intermediate-temperature ductility of low-alloy steels and has received the greatest attention by researchers. While embrittlement in this temperature range does not occur in high-purity iron,[30,37,49] it has been found in both plain-carbon steel[45] and an Fe-0.24 Si binary alloy.[66] These observations suggest that embrittlement is not possible without some precipitates, and reveal the importance of even minor amounts of residual elements.

One of the most influential elements affecting ductility in this region is aluminum. As shown in Figure 5, increasing dissolved Al content, or "ASA" (acid soluble aluminum), within the range of .02 to .06 percent causes a marked drop in hot ductility, particularly below 900 °C.[24,38,43,45,50] It also extends the upper limit of the ductility trough occuring in plain C steels to higher temperatures.[24,29,43,50] Further increases in ASA above .07 percent recovers the ductility somewhat, presumably due to AlN precipitate coarsening.[18,24,29] The action of Al in determining ductility is undoubtedly due to the preferential precipitation of AlN at the austenite grain boundaries.[44] It also refines the austenite grain size and retards austenite recrystallization. Mintz and Arrowsmith[29,45] report that increasing ASA also aggravates the effect of Nb(C,N) precipitates, causing them to become finer, more closely spaced and concentrated at the grain boundaries.

The influence of niobium is quite similar to that of aluminum, both in effect and severity as shown in Figure 6. Increasing Nb content again causes a drop in ductility values,[29,31,43,45] but it is even more influential than Al in extending the trough to higher temperatures.[29,31,37,38,43-45]

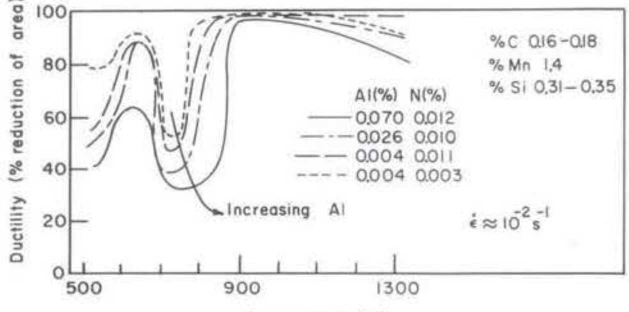

FIG. 5. Effect of aluminum on the hot ductility of steel.[43]

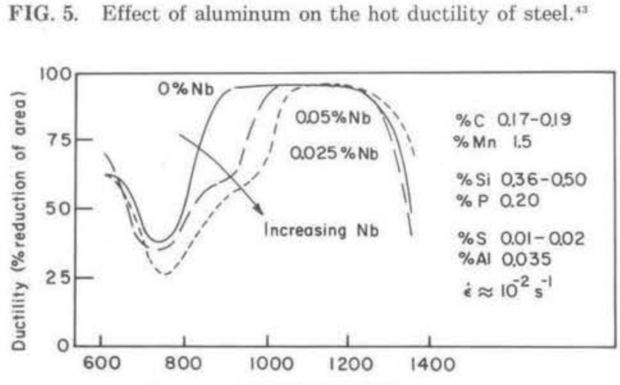

FIG. 6. Effect of niobium on the hot ductility of steel.[43]

Researchers studying Al steels both with and without Nb observe that Nb(C,N) precipitates tend to predominate at higher temperatures while AlN is more associated with the lower temperature 700-900 °C range.[38,43,45] Steels containing both Al and Nb have the deepest, widest ductility troughs. Niobium precipitates as $NbC_{.6}N_{.25}$ in high N steels,[38,64] or $NbC_{.75}$ in low N steels,[65] and is rate controlled by diffusion of Nb in austenite.[67]

In steels where boron is present, similar observations to those witnessed for Al and Nb are reported[23,46] (Figure 7). This is presumably due to the same mechanism with BN precipitates taking the place of or acting simultaneously with AlN and Nb(C,N).

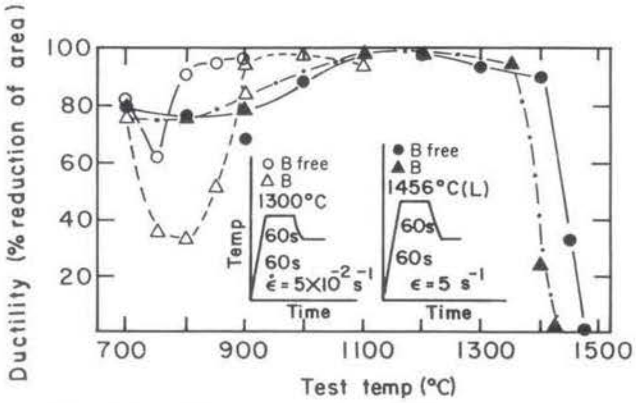

FIG. 7. Effect of boron on the hot ductility of steel.[46]

Knowing that nitride precipitates (Al and Nb) are largely responsible for lowering ductility, it is not surprising that increasing nitrogen contents are associated with decreasing ductility[24,30,31,38,45,50] and extension of the low-ductility trough to higher temperatures.[30] Its effect is not nearly as dramatic as that of Al and Nb, but unlike these elements, N is also deleterious to the low temperature properties of steel: strength and toughness.[68] Its effect is illustrated in Figure 8.

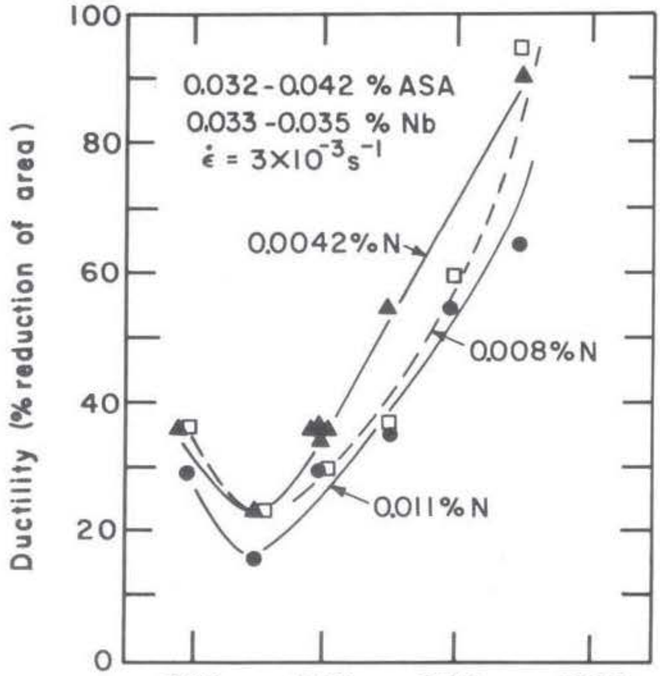

FIG. 8. Effect of nitrogen on the hot ductility of C-Mn-Al-Nb steels.[45]

Vanadium, another common micro-alloy and nitride former, is not nearly as detrimental to ductility as Al and Nb and may even be beneficial. Some believe V acts in a similar manner to Nb but has a reduced effect because of the higher solubility of VN in austenite.[31,45] Others claim that V may improve ductility,[2,13,20] particularly at lower temperatures since it hinders AlN precipitation.[69]

The final nitride former, titanium, is unique in being the only element that is unquestionably beneficial in reducing the ductility problem. Figure 9 shows that although Ti is not effective in totally removing the ductility trough,[38] it can reduce it to the same depth and extent as that observed for plain C/Mn steels.[42,45] Titanium appears able to eliminate the detrimental effects of Al[7,45] by preventing AlN formation. It does this by preferentially combining with the available nitrogen and precipitating coarser, less harmful, TiN precipitates[38,41,42,45] distributed throughout the matrix.[7] It also precipitates at higher temperatures due to its lower solubility, leaving less N for the subsequent precipitation of more detrimental nitrides. Titanium has a similar but reduced effect on Nb-bearing steels,[45] probably because Nb can also form carbides.[38]

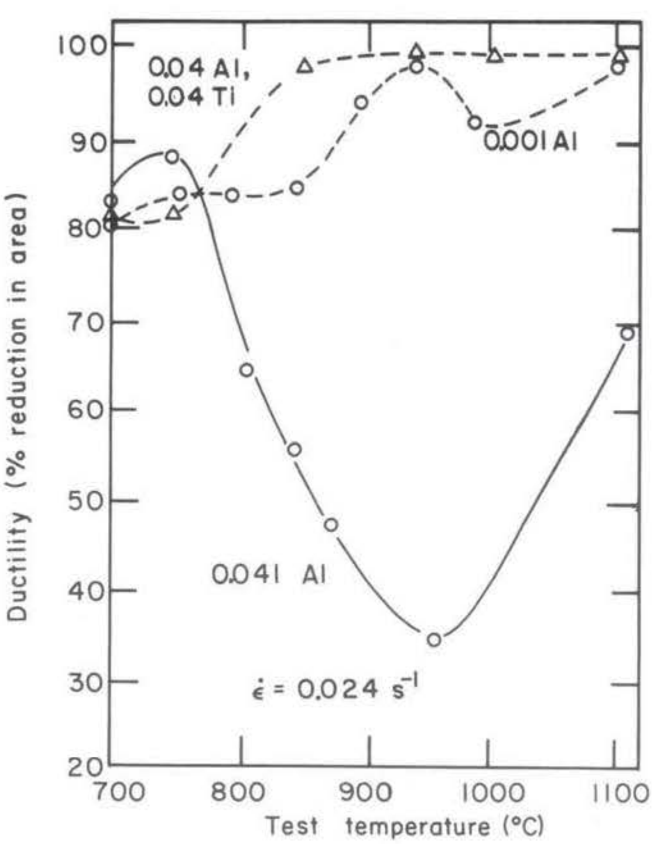

FIG. 9. Effect of titanium on the hot ductility of steel (from Funnel[42]).

Apart from nitrides, which play their most important role at lower temperatures, sulfides are seen to be detrimental over the entire range of reduced ductility. They are particularly damaging at the higher temperature 1000-1200 °C range and at higher strain rates (above 10^{-3} s^{-1}). Increasing sulfur content both deepens[32,34,37,64] and widens[30,33,37] the hot-ductility trough, as shown in Figure 10. The addition of manganese to achieve Mn/S ratios greater than 20 greatly reduces the ductility trough[30,32,34,36,37] by the same mechanism that it alleviates the hot tearing problem. However, very high manganese levels (> 1.6 percent) are reported to lower ductility, possibly due to matrix hardening.[38,51]

Because the important effects of these elements are themselves so complicated, it is difficult to determine the possible influences of other minor elements. While P is clearly detrimental at hot tearing temperatures, it is much less important in the intermediate-temperature region. Some state it has a negligible effect[32,38] while others even report a slight increase in ductility with increasing P up to 0.3 percent.[29,30,45] This was attributed to hindrance of Nb(C,N) precipitation by P segregation to the austenite grain boundaries.

Researchers hold mixed views on the influence of carbon, ranging from low-carbon steel having lower ductility[31] to medium-carbon steel being worse,[18,32,49] to there being no effect at all.[36,38] Although its action is also unclear, molybdenum additions may be beneficial. Calcium additions may improve ductility by reducing sulfur and oxygen levels.[70] Finally, oxygen has been seen to have only a slight deleterious effect,[28,54] presumably due to its contribution to (Fe, Mn, Al) inclusions and reduction in internal cleanliness. Oxide precipitates are far less damaging than either nitrides or sulfides.

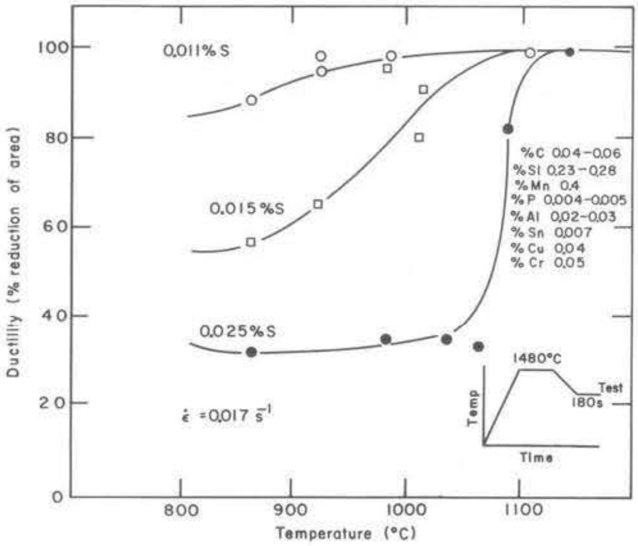

FIG. 10. Effect of sulfur on hot ductility of low-carbon steel (adapted from Weinberg[33]).

Effect of Thermal History

The other major variable affecting hot ductility is thermal history. Since time is one of the basic parameters controlling the embrittling processes, the effects of thermal history are naturally linked to strain rate. Thermal history effects are reversible since several researchers have observed that reaustenitizing a sample exhibiting low ductility at a particular test temperature, followed by a "favorable" thermal treatment and retesting at the same temperature, restores good ductility.[16,17,38] Unfortunately, there is wide disagreement as to what constitutes a deleterious or favorable thermal treatment. The findings seem to depend, at least in part, on the precipitate species responsible for embrittlement.

The time and temperature of annealing used to "initialize" the sample has been found by several researchers to be very important. More severe embrittlement and extension of the ductility trough to higher temperatures occurs when the maximum heating temperature is above the incipient grain boundary melting temperature.[33,37] The same extended embrittlement was found when annealing for short times (60 s) at 1425 °C[66] and to a lesser extent at temperatures between 1300 °C and 1400 °C.[34,37] However, Wray also reports that annealing for long times at high temperature produces extensive grain growth with no embrittlement when subsequently tested at 950 °C.[66]

Several studies report that annealing temperature has no effect on ductility but provided annealing is done above the solution temperature of the various nitrides present for sufficient time to ensure redissolution.[38,43] This should take only 300 s at 1300 °C for AlN which is quick to dissolve when reheated above its solution temperature. However, longer times in excess of 1800 s may be needed for Nb(C,N) which is much slower to dissolve.[44,71] Since many workers did not use a Gleeble apparatus, it remains uncertain whether their experimental results reflect those of actual casting conditions, where the steel is initially molten. Substantially different behavior is found if AlN precipitates still remain after annealing.[40]

Embrittlement in this temperature region is sensitive not only to annealing temperature but also to the subsequent thermal history. Slower cooling rates consistently are found to be beneficial, both when sulfides were involved[30,32,37] (.05 vs. 5 °C/s) and when AlN was the major embrittling species[24,29] (1 vs. 25 °C/s).

Many workers have found that short, (200-3000 s) isothermal periods before testing also greatly improve ductility at temperatures above 900 °C, when sulfide embrittlement is involved.[32,34,37] Figure 11 (A) illustrates this effect. However, when nitrides are responsible, holding before testing from 900 to 1800 s results in substantially reduced ductility below 1000 °C.[38,44] Even at higher temperatures, ductility decreases unless holding is done for much longer times[38,44] (more than 5400 s at 1150 °C).[42] This reflects the slow kinetics of nitride precipitation, particularly Nb(C,N).

Cooling below the A_1 temperature and reheating to the austenitic range completely eliminates embrittlement, at least when MnS is involved.[32,34,37] However, cooling either

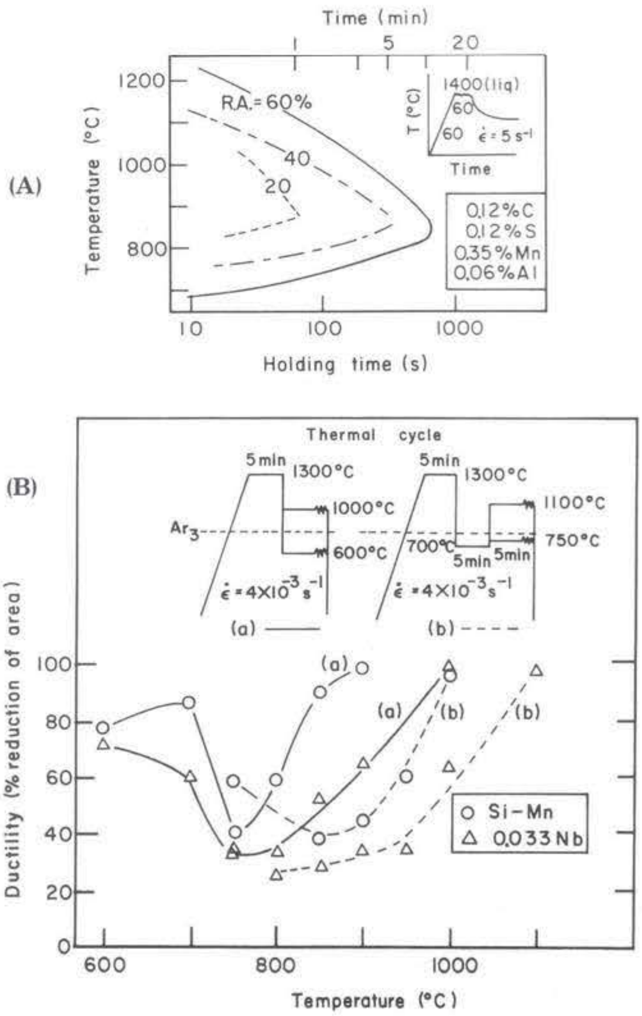

FIG. 11. Effect of thermal treatment on intermediate-temperature ductility illustrating the effects of:
(A) hold time prior to testing.[37]
(B) dropping temperature into the two-phase region prior to testing.[38]

into the two-phase region or just above the Ar_3 results in more severe nitride embrittlement. As shown in Figure 11 (B), this extends the zone of lower ductility to higher temperature, particularly when Nb is present.[38]

Finally, cycling the temperature across the A_3 temperature, such as occurs during continuous casting, was found to be very detrimental.[31,42]

Effect of Strain Rate

Researchers who test at low strain rate and/or attribute ductility losses to nitride precipitates, unanimously concur that ductility in this temperature range decreases with decreasing strain rate. Figure 12 shows that as strain rate is lowered below about 10^{-3} s^{-1}, any observed ductility trough deepens drastically. This effect is most prominent at temperatures below 900°C.[29-31,38,40,43] While lowering the strain rate also has the effect of extending the ductility trough to higher temperatures,[72] the temperature of lowest ductility remains just above the Ar_3.[38]

At higher strain rates (above 10^{-3} s^{-1}), the influence of this variable is not as clear. Researchers attributing embrittlement to nitride (Al or Nb) precipitates still find lower strain rates consistently detrimental to ductility.[37,40,66] However, many of those attributing embrittlement to sulfide (Fe or Mn) precipitates find decreasing strain rate either has no influence[32] or it increases ductility,[37] particularly at higher temperatures (900-1200°C).[38]

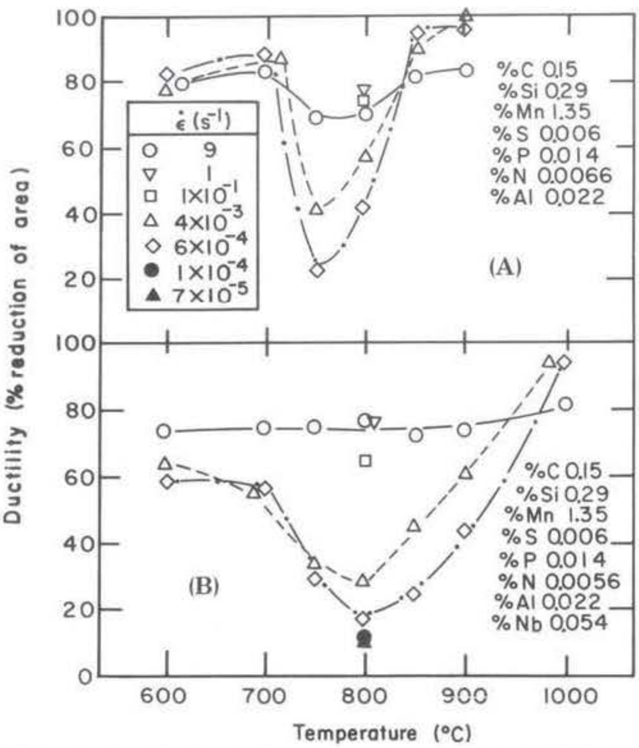

FIG. 12. Dependence of intermediate-temperature ductility on strain rate[38] for:
(A) low-carbon, Si-Mn steel
(B) Nb-bearing steel

Effect of Grain Size

The grain boundaries are the weak link in steel (and other metals) at elevated temperatures. Thus, coarse-grained materials should exhibit lower ductility, particularly at lower strain rates, where the grain-boundary weakening mechanisms have time to operate. This is because strain concentration at the weakened grain boundaries is enhanced when less grain boundary area is present. Many studies indeed have determined coarse grain size to be associated with lower ductility,[30,31,37,39,43,45] but several others found fine grains to be worse[33,41,42] and another stated it was not important.[38] These contradictory findings are more understandable when one considers that grain size is intrinsically related to other variables such as grain refining agents and thermal history which themselves are highly influential on hot ductility.

The effect of Al additions on grain size is particularly important to note. Some of the aluminum added to steel acts as a deoxidant, being the most effective and economical element to perform this role. More importantly, however, the remainder is dissolved in the steel and is very effective in controlling the austenitic grain size. The dissolved aluminum, or other grain refining element such as Nb, Ti and to some extent Zr and V, accomplishes this by preventing the grain growth that normally occurs at high temperature with increasing time. These alloying elements act by forming "obstruction agents" which mechanically obstruct grain growth by pinning the austenite grain boundaries. The obstruction agents are fine, discrete particles (usually carbides or nitrides) that precipitate during cooling at the austenite grain boundaries when their solubility limit has been exceeded. AlN precipitates are particularly effective,[73] but niobium carbonitrides and titanium nitrides are also suitable.[74]

With increasing temperature during reheating, the fine particles both coalesce (Ostwald ripening) and start to dissolve back into solid solution. As the precipitates coarsen and reduce in number, the pinning effect is reduced. When the rate of release of energy per unit displacement of grain boundary (during grain growth) exceeds the rate of increase in energy due to the unpinning process, grain growth occurs.[75] At this critical temperature, called the grain coarsening temperature, detrimental secondary recrystallization can begin. It is manifested by the rapid growth of a few grains to large sizes which lowers final product quality and consistency.

Grain size control is, therefore, very desirable during later reheating stages, prior to rolling. The fine grain size resulting from Al addition has been correlated with a decrease in low-temperature ductile/brittle transition temperature, an increase in both low-temperature strength and toughness and improved weldability, aging resistance and distortion resistance.[18]

However, this mechanism by which Al prevents grain coarsening is the same one that contributes to a lowered ductility in the intermediate-temperature range. Indeed, the same factors that result in lowered ductility have been found to achieve the finest grain size. For example, the grain coarsening temperature of steel depends to a large extent on the solution temperature of AlN.[18,76] The highest grain coarsening temperatures result from large volume fractions of very fine precipitates made from intermetallic compounds with low solubility products.[76] Because of its low solubility (Table I) and its readiness to form very fine precipitates (< 1 micron), AlN is a very effective obstruction agent. However, if the aluminum content of the steel is too high (above 0.08 percent), AlN can precipitate at higher temperatures as coarse particles which reduce its

TABLE I - SOLUBILITY OF VARIOUS NITRIDES IN AUSTENITE[77]

Compound	Solubility (weight percent)	
VN	$\log_{10}$ [V] [N] = −8300/T (°K) + 3.46	very soluble
NbN	$\log_{10}$ [Nb] [N] = −8500/T + 2.80	↓
AlN	$\log_{10}$ [Al] [N] = −6770/T + 1.03	↓
TiN	$\log_{10}$ [Ti] [N] ≅ −7.0 @ 1100°C	↓
ZrN		very stable

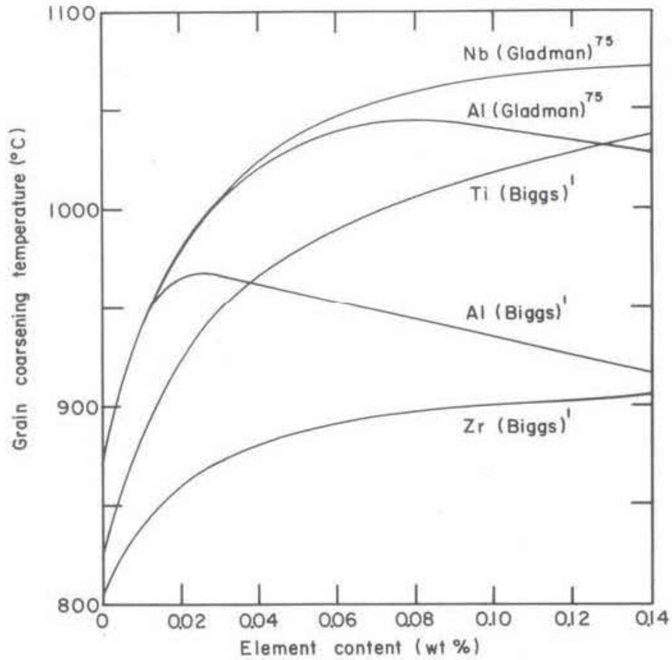

FIG. 13. Effect of various elements on the grain-coarsening temperature of plain-carbon steel.

effectiveness and results in a lower grain coarsening temperature, as shown in Figure 13. An optimum range exists at 0.015 - 0.05 percent Al.[3] The addition of titanium will preferentially combine with the nitrogen to reduce AlN formation and again lower the grain coarsening temperature.

Thus, when a fine grain size is observed associated with a lower ductility, it may simply reflect the effectiveness of nitride precipitates both in preventing grain growth and in weakening the austenite grain boundaries of that particular sample.[42] On the other hand, the action of AlN precipitates may prevent dynamic recrystallization which could result in a coarser grain size. In this case, lower ductility would appear associated with the coarse grain size for the additional reasons that AlN precipitate action has accelerated grain boundary cavity nucleation[39] and reduced grain boundary mobility.[45] In conclusion, grain size itself is not always a major factor controlling the hot ductility of steel. It often simply reflects the influence of the processes which do control ductility such as the effectiveness of nitride precipitates in grain boundary pinning.

Mechanism

It is fairly well agreed that trends observed in the intermediate-temperature zone of reduced ductility can be explained largely by the action of precipitates at the austenite grain boundaries. However, theories differ as to how these precipitates operate to reduce ductility. While many of the apparent contradictory findings can be explained, the phenomena are far from being completely understood. There appear to be at least three separate mechanisms operating both simultaneously and independently to account for the complex behavior observed in this temperature region. Under any particular set of conditions, any one of these is likely to predominate.

Wray has represented these different fracture mechanisms in terms of different strain-rate temperature zones via a fracture map shown in Figure 14.[66] Zone A is the high-temperature zone responsible for hot tearing. Zones B, C and D correspond to three mechanisms involving the austenitic phase: sulfide embrittlement at high strain rate, ductile intragranular fracture, and intergranular creep fracture, respectively, which operate in this intermediate-temperature zone. Zone E refers to the embrittling mechanism operating in the two-phase austenite and ferrite region below the A_3 temperature. These same labels were used to identify the low-ductility zones shown in Figure 3.

At low strain rates (below 10^{-3} s^{-1}), embrittlement occurs in austenite by the nucleation, growth and coalescence of grain boundary voids. Under stress, small precipitate particles assist in initiating micro-fissures, or equiaxed, creep cavities with faceted surfaces. With increasing strain, the creep cavity density increases.[78] In addition, growth occurs by spreading along the grain boundary plane.[39] If either the rate of cavity nucleation is high,[78] or extensive cavity growth or coalescence can occur along the boundaries, then low ductility, intergranular fracture results. The final morphology of the creep cavities changes with increasing strain rate (or stress) from lens-shaped to wedge-shaped, accompanied by more plastic flow.[72] At high strain rates, voids have insufficient time to nucleate and transgranular rupture occurs, in the absence of other embrittling phenomena.

The work of Funnell[41,42] and others[29,37,40] suggests a mechanism for precipitate action in this intermediate-temperature region at low strain rate. If grain boundaries can migrate away from developing cavities, then cavity growth stops, stress concentration at the grain boundary is relaxed, and ductility is preserved. However, in addition to providing initiation sites for void nucleation, precipitates also tend to hinder or prevent grain boundary mobility.[40-42] This encourages cavity coalescence and intergranular failure. Alternatively, the action of precipitates may enhance grain boundary sliding.[37,41,42] According to this mechanism, the voids nucleate, grow, and coalesce along the austenite grain boundaries by the relative movement of neighboring grains along their boundaries.

In either case, the thermal-history strain-rate combinations that produce many fine precipitates at the austenite grain boundaries result in lowest ductility.[40-42,45] Mintz and Arrowsmith[45] suggest that a critical particle size exists, above which grain boundary migration can occur. This size would be a function of particle volume fraction, initial grain size and the stored energy of deformation. Only a sufficiently large number of precipitates smaller than this critical size causes grain boundary pinning and the resultant loss of ductility. Thus, ductility decreases for decreasing precipitate size and increasing volume fraction.

Nitrides are the principal precipitates responsible for enhancing this mechanism. This is because AlN and Nb(C,N) are slow to nucleate in austenite[29,40,64,71,79] and under average cooling rates, produce very fine precipitates, averaging about 100 nm in diameter for AlN[29,41] and smaller than 50 nm for Nb(C,N).[29,38] Because sulfide precipitates tend to be much larger (200-5000 nm),[54] as well as to nucleate easily and grow rapidly, they are much less effective at grain boundary pinning by this mechanism.

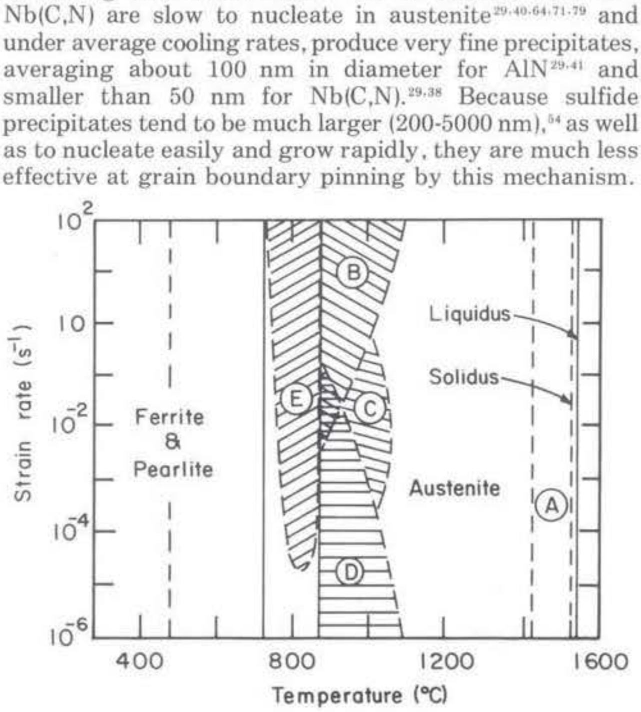

FIG. 14. Possible fracture zones mapped for 0.2 percent C plain-carbon steel in strain-rate temperature space.[66]

An additional action of Nb(C,N) precipitates in enhancing embrittlement in this zone is the concentration of strain at the austenite grain boundaries. This occurs when networks of fine, intragranular Nb(C,N) particles cause precipitate hardening in the matrix. Many of the previously discussed observations can be explained by this mechanism. A low strain rate allows time for the diffusion-controlled processes of (Al, Nb, B) nitride precipitation and grain boundary void coalescence to take effect. As strain rate is decreased, these embrittling mechanisms are enhanced. Higher Al, Nb, B or N levels increase nitride precipitate solubility products which result in increased precipitation rates.[68] Fast cooling rates or cooling to lower temperatures again encourages precipitation[29,38,71] and results in finer, more closely spaced particles.[29] This is because solubility products decrease logarithmically with decreasing temperature (Table I) so supersaturation, the driving force for nucleation, is increased. Cooling to lower temperatures and reheating before testing promotes rapid nitride precipitation for the same reason. Any of these factors causing increased nitride precipitation rates tends to lower ductility and extend the ductility trough to higher temperatures.

While usually associated with strain rates below 10^{-3} s^{-1}, grain boundary sliding has been found to be at least partially responsible for embrittlement at strain rates as high as 10^{-1} s^{-1} in a 0.54 Nb steel at 900°C.[38] However, Ouchi and Matusumoto[38] remark that because austenite grain boundary sliding was not sensitive to Nb, or N content or increasing temperature above 900°C, it cannot be the controlling factor for embrittlement.

The predominant mechanism embrittling steel at high strain rates involves (Fe, Mn) sulfide precipitates. Sulfur strongly and rapidly segregates to the austenite grain boundaries to form weak sulfide films which can fail in a manner reminiscent of high-temperature or hot-tearing embrittlement.[32,33] Indeed, liquid-film failure itself is possible at temperatures above the Fe-FeS eutectic if local remelting of sulfur rich pockets can occur.[33]

Slow cooling or isothermal holding allows time for the slow diffusing Mn to combine with S and form less harmful MnS precipitates which reduces FeS formation at the grain boundaries. In addition, high Mn/S ratios encourage harmless MnS precipitation inside the grains. This also explains the beneficial effects of high Mn/S ratio or low S level. However, liquid-film failure alone cannot explain the problem since ductility losses occur even with Mn/S ratios and temperatures well above those required to prevent liquid FeS film formation. Thus, in addition, the precipitates themselves must be harmful,[34] presumably in a manner similar to that of nitrides.

The conflicting results obtained for varying thermal treatments and strain rate can be better understood by considering precipitate thermodynamics. Above 1200°C, thermally activated processes such as dislocation climb and recrystallization restore grain boundary mobility. With increasing time and temperature, precipitates both coarsen and dissolve, reducing their effectiveness at pinning grain boundaries and recovering ductility. The aggravated embrittlement caused by high annealing temperatures is presumably due to the complete dissolution of MnS precipitates that occurs at 1420°C.[66] In addition, if local grain boundary melting can occur before cooling, this detrimental action reduces ductility at lower temperatures.[33,37]

Between 1200°C and the A_3 temperature, ductility depends on the size, number and location of precipitates produced by the previous thermal treatment. Slow cooling and isothermal holding induce both precipitate nucleation and growth. The same conditions of increased time that allow large numbers of fine nitride precipitates to form and lower ductility, cause sulfide precipitates to coarsen and thereby improve ductility. Cooling below the A_1 and reheating nucleates precipitates which grow rapidly, and for the most part, harmlessly inside the grains. At higher strain rates, less time is allowed for the coarsening of sulfide precipitates, resulting in more effective grain boundary pinning and thereby reduced ductility. Because they precipitate and grow so much faster than nitrides, sulfides and oxides probably play their greatest role in the upper 900-1200°C temperature region at higher strain rates.

At intermediate strain rates, a third mechanism comes into play. This is the competition between dynamic recrystallization and plastic tensile instability. If it can occur, dynamic recrystallization completely relaxes any local stress concentrations and creates fresh grain boundaries that trap harmful precipitates and voids inside the grains. This results in much improved ductility. In austenite above 1050°C, this occurs so readily that high ductility is usually assured. However, if the Considere strain[66] (the strain at neck formation or when $d\sigma/d\varepsilon$ equals σ) is less than the strain required for dynamic recrystallization, then premature necking and failure occurs. Thus, factors which either retard dynamic recrystallization or lower the work hardening parameter reduce ductility. Both AlN and Nb(C,N) retard austenite recrystallization[37,43-45] which explains further the detrimental effects of nitrides on ductility. Norstrom[80] explains the different observed effects of strain rate on ductility by this mechanism, which is illustrated schematically in Figure 15. At a lower strain rate the recovery process is active, which reduces the driving force for dynamic recrystallization and thereby delays the ductility-improved recrystallization process. At high strain rate, the lack of time again prevents dynamic recrystallization, possibly leading to an early transgranular, ductile fracture. Thus, the best ductility should occur at intermediate strain rates, especially if the temperature is high enough to allow recrystallization. The absolute magnitude of the optimum strain rate will depend on the creep, recovery and recrystallization characteristics which depend on steel composition.

Since grain boundary pinning processes become less effective with increasing strain rate, the prevention of recrystallization may be the main process responsible for continued low ductility at intermediate strain rates. However, Ouchi and Matusumoto[38] point out that this mechanism clearly is not the only one operating in the intermediate-temperature region since it cannot explain:

(1) The occurrence of fracture with less than the Considere strain.
(2) The importance of thermal history.
(3) Why increasing strain rate sometimes improves ductility while it retards dynamic recrystallization, as well as increasing the Considere strain.
(4) Why Nb consistently lowers ductility at lower strain rates while it suppresses recovery as well as recrystallization.

Clearly, the mechanical behavior of steel in this intermediate-temperature region is not fully understood. Further work needs to be done to unravel the complexities of the different embrittling mechanisms.

LOW-TEMPERATURE ZONE

The third zone of low ductility in steel occurs in the two-phase austenite-ferrite region below the Ar_3 temperature. It corresponds to Zone E in Figures 3 and 14. In many respects, it is a continuation of the previously discussed intermediate-temperature zone involving single-phase austenite. However, the presence of ferrite appears to involve another embrittling mechanism in the temperature range of 600-900°C.

The fracture resulting from tests done in the lower temperature, or "two-phase" zone, although appearing brittle due to its intergranular nature and lack of macroscopic strain-to-fracture, is thought to be a ductile failure at the austenite grain boundaries on a microscopic scale.[38] The fracture surface was covered with dimples, many of which contained a precipitate particle.[37,38,50] These precipitates consisted mainly of AlN[18,24,37,43,44,50] but other nitrides, sulfides and a few oxides were also found.[30,37] In steels containing Nb or B, Nb(C,N)[29,37,38,44,47] and BN[23,46] precipitates often were found, both on the fracture surface at the prior austenitic grain boundaries and within the matrix.

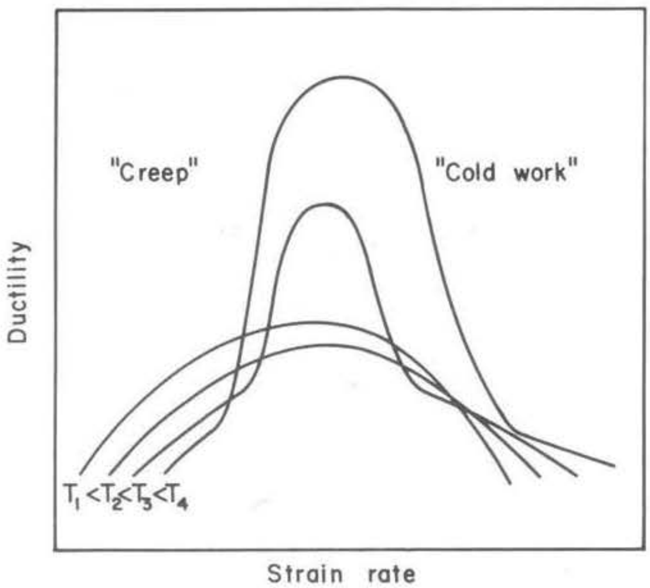

FIG. 15. Schematic representation of the effects of strain rate and temperature on the hot ductility of steel.[80]

The same detrimental effects of nitride-precipitate forming elements were found in this two-phase zone.[38] Increased Mn or Si also decreases ductility slightly below 750°C.[38] In addition, residual (Cu, Sn, Sb, As) and impurity (S, P) elements segregate to the ferrite grain boundaries to further lower ductility.[25]

The effects of thermal treatment also are generally the same except that increased holding time at 750°C was found to improve ductility,[38] as shown in Figure 16. Many researchers observed that the temperature at the start of transformation was associated with the minimum ductility and that ductility rapidly improved with decreasing temperature below the Ar_3.[34,37,43,44] Most researchers agree that lowering strain rate in the two-phase zone drastically reduces ductility,[37] particularly near the A_3 temperature.[37,38] However, Wray[66] suggests that embrittlement by this mechanism may decline at low strain rate as shown in Zone E in Figure 14.

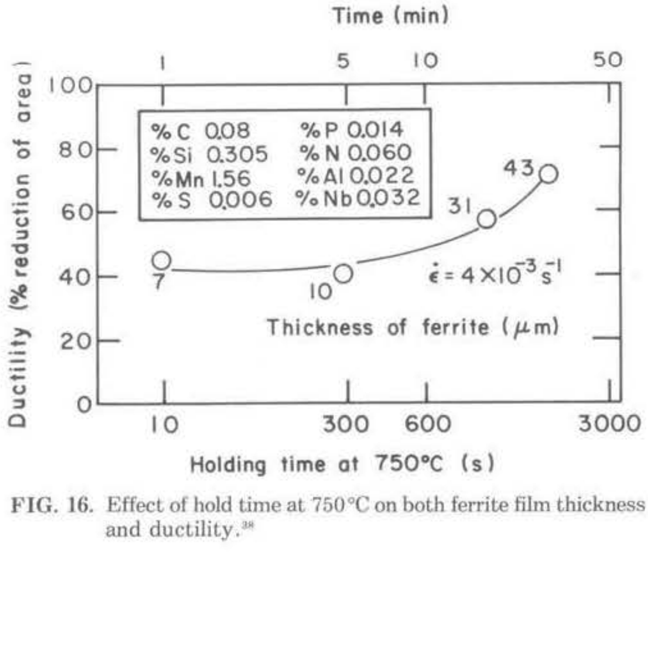

FIG. 16. Effect of hold time at 750°C on both ferrite film thickness and ductility.[38]

Mechanisms

The additional mechanism that has been proposed to control embrittlement in the two-phase zone places only secondary importance on the action of precipitates.[7,30,38,47,50] Grain boundary weakness is instead attributed mainly to strain concentration at the primary ferrite film forming along the austenite grain boundaries. This occurs because, at the same temperature, ferrite is more ductile[35,48] and has less strength[66] than austenite. This is due partly to the higher atomic diffusivity of ferrite and to the larger number of slip systems in bcc (48) compared with fcc atomic structures (12).[48] The austenite matrix can be hardened further by the addition of elements such as Cr[23] or by intragranular precipitation[34] such as Nb(C,N).[25,29,44]

The presence of precipitates, particularly nitrides, further exacerbates the problem by enhancing strain concentration and embrittling the grain-boundary ferrite, each precipitate nucleating a microvoid.[50,51] In addition, the primary ferrite encourages preferential precipitation at the grain boundaries because nitrides have a much lower solubility in ferrite than in austenite.[44,71] For example, Figure 17 shows that AlN precipitation, which can take several hours in single-phase austenite,[79] occurs within minutes once ferrite is present.[71]

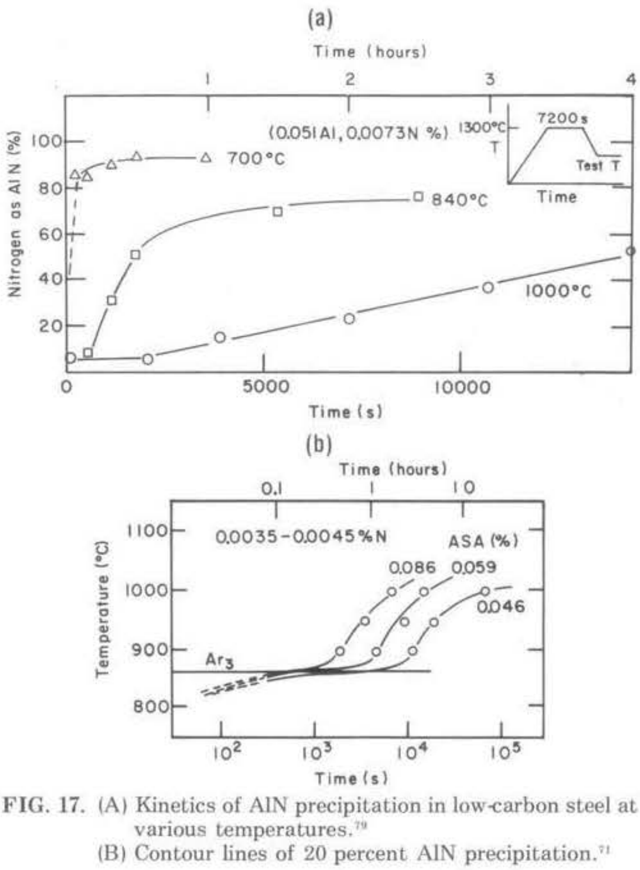

FIG. 17. (A) Kinetics of AlN precipitation in low-carbon steel at various temperatures.[79]
(B) Contour lines of 20 percent AlN precipitation.[71]

With continued stress, the microvoids multiply and the result is an intergranular, but microscopically, ductile fracture. Ductility is at a minimum when the pockets of nucleating primary ferrite first link into a continuous film at the austenite grain boundaries. The thickness of this proeutectoid, ferrite film is the controlling factor for ductility according to this mechanism.[37] With lower temperatures or longer holding times, the accompanying increased thickness of the ferrite film (Figure 16) is believed to be responsible for the observed improvement in ductility.[38] Yamanaka[50] successfully correlated the ductility trough with the theoretical fracture strain calculated from the equations of Gurland and Plateau[81] describing ductile fracture around inclusions.

This mechanism, illustrated in Figure 18, is consistent with several of the previously discussed observations. It explains the dimpled, ductile, intergranular appearance of the fracture surface and the association of minimum ductility with the Ar_3 temperature. Increasing both grain boundary and intragranular precipitates, particularly nitrides, naturally lowers ductility. The enhanced precipitation rate of nitrides in ferrite also helps to explain the decrease in the ductility of austenite when cooling into the two-phase region and reheating.[38] Ouchi and Matusumoto[38] suggest that increasing strain rate might suppress strain concentration, thereby increasing ductility. However, strain concentration may occur even at high strain rate, as evidenced by the ductility drop in the two-phase region of Armco iron while testing at $0.47\ s^{-1}$.[48]

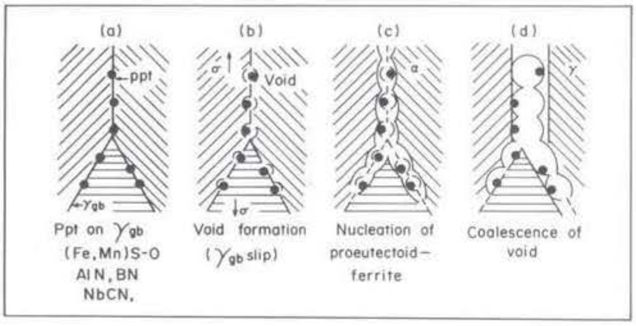

FIG. 18. Mechanism for embrittlement in the low-temperature or two-phase zone.[37]

Many researchers feel that strain concentration in the ferrite is not necessary for embrittlement in this zone.[37,41] The presence of a ductility drop even in a single-phase austenitic stainless steel over a similar temperature region is evidence of this.[37,41,66] The continued loss in ductility below the Ar_3 may simply reflect the continued operation of embrittling mechanisms from the intermediate-temperature region such as grain boundary sliding. In many studies, the worst ductility was found at 750 °C, regardless of whether the Ar_3 was at 750 °C or lower. This implies that ductility losses still may be controlled by thermally activated processes involving precipitates even if primary ferrite is present.[37] The improvement in ductility with continuing temperature reduction then would be due to the gradual decline in importance of these mechanisms.[44] Wray states that the phase-transformation process may improve an inherent weakness in austenite by trapping harmful precipitates and voids inside new grains.[66] Below the A_1 temperature, this surely is the case. The new, fully-transformed, fine-grained, ferrite and pearlite structure generally has excellent ductility.[38]

However, a possible exception to this exists for higher carbon steels where a fourth low-ductility zone may come into play if the cooling conditions and composition are such that a permanent, thin, ferrite network results in embrittlement below the A_1 temperature.[3] Carbon contents near 0.8 percent would be the most susceptible, but alloying elements such as Cr and Mn may also influence the amount of primary ferrite that forms and produce large volume fractions of pearlite at lower carbon contents. This mechanism has been documented for hypereutectoid steels over a wide range of temperatures where cementite is the brittle, grain-boundary phase.[66] Wray suggests that this fracture zone begins at lower temperatures with decreasing carbon content for the ferrite-pearlite case.[66] This is represented schematically in Zone F in Figure 19, which shows the relationship with carbon for Zones A-E as well. Indeed, several researchers studying medium-carbon steels found a drop in ductility between 600 and 700 °C that was also associated with high aluminum contents.[3,16,17] Titanium additions were found to be beneficial[1,3,7,16] presumably acting by the same mechanism that improves intermediate temperature ductility. In addition, Zr additions[30] and possibly V as well, were found to alleviate the problem, although not as effectively.[3,16,17] Higher N levels were also found to be detrimental if accompanied by aluminum.[1,3]

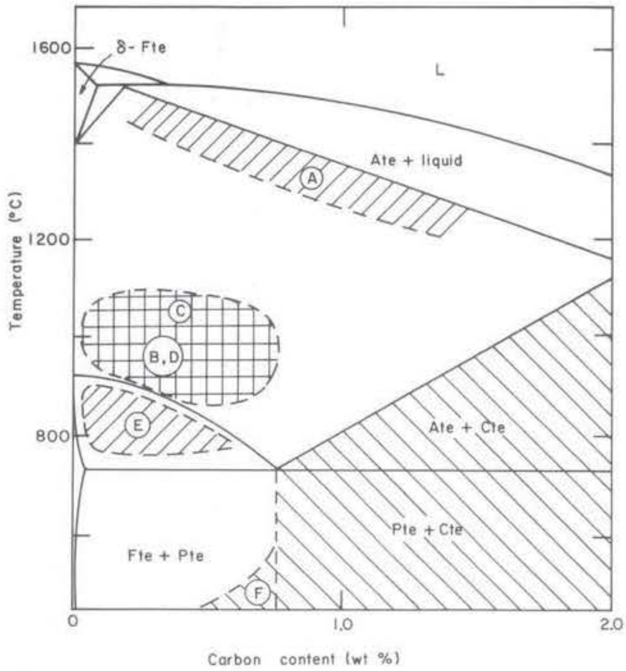

FIG. 19. Schematic diagram showing the zones of embrittlement (Figure 14) at intermediate strain rates for the Fe-C system.[66]

IMPLICATIONS OF DUCTILITY FOR PANEL CRACKING

The previous discussion of the zones of lowered ductility affecting steel has several important implications for the panel cracking problem. Since thermal cracking can be prevented if the material can accommodate as little as 2 percent strain, then a severe ductility loss must be encountered before panel cracking can occur. The different zones of reduced ductility are manifested under different and sometimes opposite processing conditions. It is, therefore, imperative to identify the particular embrittling mechanism(s) responsible if solutions to panel cracking are to be found.

The cooling rates of a large ingot solidfying in its mold or even during air cooling are very low. This results in strain rates, due to thermal contraction, on the order of $2 \times 10^{-6}\ s^{-1}$ or less.[7] Relative to the previously discussed ductility studies, this strain rate is extremely low. This fact and the importance of AlN together rule out several embrittling mechanisms from being responsible for panel cracking. The first of these is the high-temperature, hot-tearing zone of embrittlement. Panel cracking cannot be due to this mechanism for several additional reasons. It affects steels even with high Mn/S ratios and low S contents and the effects of subsequent thermal history are too important. Finally, it is not responsive to changes in initial casting conditions.

The sulfide embrittling mechanism affecting intermediate-temperature austenite is also a very unlikely contributor to panel cracking. Besides the high Mn/S ratios (usually greater than 50) and low S levels, the slow cooling rates and low strain rates encountered in ingot casting would undoubtedly coarsen sulfide precipitates and eliminate embrittlment by this mechanism.

Since the total strain rates involved in panel cracking are very low and the cracks are intergranular, the effects of recrystallization and tensile plastic instability are also unimportant.

This leads to the conclusion that the embrittling mechanism responsible for panel cracking must be either grain boundary void coalescence in austenite just above the A_3 temperature or the lower temperature zone of embrittlement. In either case, improving the ductility of affected grades near the A_3 temperature would be beneficial.

One method of achieving this is through alteration of the steel composition to prevent the formation of detrimental, fine nitride precipitates. This can be achieved most effectively by lowering the addition of nitride-forming elements such as Al, Nb, B and especially N. Alternatively, raising the ASA level markedly[24,42] or adding Ti would produce coarse or harmless precipitates and again improve ductility. Finally, lowering S and O levels or adding Ca or Mn would only help to reduce sulfide embrittlement, but at least would not do any harm.

A second solution to avoid the production of fine, nitride precipitates is through alteration of the thermal treatment. Unfortunately, the effects of thermal history are still in some dispute so it is not known what thermal treatment is best. Suggestions have been made that temperature-time cycles to either coarsen the precipitates[30,37,42] or keep them in solution[42] will both alleviate the low-ductility problem. However, the actual thermal histories to use are still unknown and the possibility of exacerbating the problem instead is quite likely. Thus, solutions to nitride embrittlement by alternate thermal treatment are not obvious given our present level of understanding.

The final solution, if it can be called that, is simply to avoid straining the steel significantly while it is in a region of low ductility. One way to achieve this might be to strip the ingot from the mold early, keeping it warm during transport, and reheating it quickly in an attempt to prevent the ingot surface from falling into the low-ductility temperature range. Then, immediately subsequent processing would be done using high strain-rate operations such as rolling or forging where ductility problems related to nitrides are less likely.[38,42] However, other embrittling mechanisms have been seen to operate at higher strain rates. In addition, AlN precipitation itself is accelerated by deformation.[29,40,44,64] Because of the wide range of temperatures affected by the ductility trough, it is virtually impossible to process steels only under conditions where good ductility exists.[42] Thus, the most practical way to apply this solution is by altering the thermal treatment to reduce the stresses acting in the solidifying and cooling ingot.

SUMMARY

The hot ductility of steel at elevated temperature has been reviewed to provide a fundamental background for the study of panel-crack formation in ingots. Three temperature ranges of reduced ductility can be discerned and two of these contribute to panel cracking. One of the zones exists in the two-phase austenite-ferrite region below the Ar_3 temperature, and is due to the concentration of strain in primary ferrite films at austenite grain boundaries. The presence of precipitates, particularly nitrides, further enhances the low ductility and fracture is intergranular. The second of the reduced-ductility temperature ranges that influences panel cracking extends from the Ar_3 temperature to as high as 1200°C. Fracture in this zone also is usually intergranular along austenite grain boundaries and the fracture surface exhibits precipitates of sulfides, oxides, nitrides and carbonitrides. A creep-type failure appears to be operative due to the coalescence of voids nucleating at the grain-boundary precipitates. Thus the kinetics of dissolution, nucleation and growth of precipitates are central to the ductility of steel in this zone. Steels containing Al and Nb, which form highly stable nitrides and carbonitrides, have the deepest ductility trough over a wide temperature range.

A third zone of low ductility is observed within 30-70°C of the solidus temperature and results from the presence of interdendritic liquid films rich in P and S. This region of reduced ductility, although a major factor in crack formation in continuous casting, does not affect panel cracking.

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