

Experience in Producing Vanadium-Microalloyed Steels By Thin-Slab-Casting Steel Technology

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ABSTRACT

Vanadium-nitrogen microalloying has proven to be highly compatible with the thin-slab casting and the direct-charging steelmaking process. Metallurgical features of vanadium and nitrogen are related to the unique process parameters associated with this process.

Key words: Thin slab, vanadium, nitrogen, solubility, precipitation, grain size, strain aging.

INTRODUCTION: GROWTH OF THIN-SLAB CASTING AND DIRECT CHARGING

The thin-slab-casting revolution has taken place over a relatively short time period, starting first in Crawfordsville, Indiana, U.S.A. in 1989. The immediate success of that operation led to many more plants being built, first in North America and then spreading across the world¹. Initially, the emphasis was on the obvious economic and ecological benefits gained by retaining part of the heat of casting in the slabs through to the rolling process, and casting to a "near net shape," reducing the amount of hot working necessary to reduce the material to final gauge. Using electric-furnace steel with most of the iron units derived from scrap, the capital costs for building these types of mills was less than half that of conventional mills using blast furnaces and basic-oxygen furnaces.

Because of the rapid industrial expansion of this new strip-manufacturing process, the effect of the significant metallurgical changes of thin-slab casting was initially managed on a real-time basis by those metallurgists left with the responsibility of providing a saleable quality product off the mill. Laboratory-based metallurgical research projects quickly followed, with a large number of industrial and academic organizations evaluating the problems and promises of the newly developed and economically successful thin-slab casting and direct-charging process. A basic understanding of the impact of these processing differences was rapidly developed through these projects.

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Because many of these new mills did not have downstream finishing processes such as cold rolling and coating, there was a natural commercial interest in developing applications for as-rolled hot band. Since much of the market for hot band was for structural and automotive applications, the demand for higher strength grades was a natural extension of this hot-band market. Conventional microalloying approaches for high-strength, low-alloy (HSLA) strip steels presented some processing complications that were new and required some alternative solutions². For many mills, the vanadium and nitrogen alloying system has proven to be highly effective and compatible with the processing peculiarities of the thin-slab-casting and direct-charging process.

PROCESSING CHARACTERISTICS: VARIATIONS FROM NORMAL PRACTICES

The thin-slab-casting, direct-charging and finish-rolling process has a number of characteristics that vary from conventional blast-furnace / BOF / slab-cast / slab-reheat / roughing-mill / finishing-mill routings. First, steelmaking often uses electric-arc-furnace (EAF) melting from predominately scrap charges, and the refining process takes place in separate ladle-furnace operations. While this is not true in all cases, it is by far the majority, especially for greenfield sites. Second, the casting process involves rapid cooling, necessary for high-volume production in thin slabs from 50 to 100 mm in thickness. Third, the slab is directly charged into a tunnel holding furnace without undergoing an austenite (γ) to ferrite (α) transformation. Fourth, the tunnel furnace is limited in reheating capabilities, typically from to 1100 to 1150°C maximum.

Finally, the 50-mm slab will enter directly into a 5- or 6-stand finishing mill without undergoing a roughing mill reduction. Also, the 80- to 100-mm slab casters will have one or two roughing mills prior to a 5-stand finishing mill. In those cases, a transfer system of coil boxes or holding furnaces will be installed after the roughing mills to equalize temperatures before entering the finishing-mill stands. This step provides a speed "break" between the start and finish rolling

speeds to allow more reasonable entry speeds in the first reduction pass. The remainder of the process, from finish rolling through a runout table usually equipped with significant accelerated-cooling capability and into a down coiling system, will be similar to existing modern strip mills. Many of the mills have the capability to approach 1-mm final thickness, although most commonly the final gauge will be 1.5-mm or larger.

MELTING CONSIDERATIONS: CONTROLLING RESIDUAL ELEMENTS

With EAF scrap-based melting, the normal issues of metallic and nitrogen residuals have to be considered. High-quality scrap, along with pure iron units like DRI and HBI, are used for melting to minimize metallic residuals. Early surface-quality problems from hot shortness were identified in some mills. Maximum copper levels for these steels were determined for each mill according to their process capabilities. Copper is generally restricted to less than 0.15% in these mills today.

Aluminum levels are controlled to levels less than 0.035%^{1,3}. Higher levels are not needed for deoxidation and can help form inclusions which can interfere with the fluid flow in the caster. Calcium treatment is also commonly used to reduce the clogging tendency of aluminum inclusions and to modify oxides and sulfides so that they remain globular through the rolling process. These modifications improve the isotropic nature of the product, particularly the formability and toughness in the transverse direction.

Sulfur control has proven to be critical for thin-slab casting. Sulfur control can be a problem for either basic-oxygen or electric-arc furnace melting, but well-established techniques are available to reduce the sulfur levels. With the EAF process, this is done both in the furnace and later in the ladle using calcium treatments described above. Each mill will have established specific control levels they believe necessary for casting a quality product. Maximum sulfur levels are generally below 0.01%³.

Nitrogen levels are characteristically quite high from EAF's when producing a low-carbon melt (below peritectic) preferred for the thin-slab casters. While many thin-slab-casting operations will claim to have the capability to produce steels in the peritectic region (nominally from 0.08 to 0.16% C), most would prefer not to be casting in this carbon range. Staying below the peritectic while casting requires restricted tapping carbon levels, particularly for high-strength steels where high-manganese levels are required.

The carbon pickup from the ferromanganese, as well as other alloys, must be accounted for at the time of tapping. Melt-in carbon levels are kept low to minimize time required to finish the decarburization process. Nitrogen can be removed from the steel bath by absorption in the CO bubbles formed during the decarburization process. Because the amount of decarburization is limited in a low-carbon bath, the nitrogen introduced from the scrap charge and the electric-arc is not easily removed. Therefore, nitrogen levels remain relatively high.

While blast-furnace and basic-oxygen-furnace melting may routinely result in nitrogen levels from 30 to 50 ppm, electric-arc-furnace melting may average from 70 to 100 ppm. Even though some mills have developed the capability to produce EAF steels at substantially lower nitrogen levels using selected raw materials and special processing, nitrogen levels continue to be a major difference between conventional BOF-melted and EAF-melted thin-slab strip steels. Even with the use of aluminum, considerable free nitrogen will remain in the hot-band steel.

Additions of the microalloying elements -- vanadium, niobium and titanium -- all eventually combine (precipitate) with the interstitial elements, carbon and nitrogen. The effect of these microalloys on the steel properties is to a great extent determined by the type and timing of that precipitation. For this reason, when microalloys are added to steel, nitrogen becomes an integral part of the alloy system as much as carbon.

Significant differences in residual nitrogen will play a large role in defining the performance of a given microalloy addition. Managing the nitrogen level therefore becomes a necessary part of producing consistent microalloyed-steel properties. Because of the higher nitrogen levels inherent in electric-arc-furnace steels, the microalloy used should be compatible with the additional nitrogen.

CASTING CONSIDERATIONS: RAPID SOLIDIFICATION

From a physical-metallurgy standpoint, the defining characteristic of the thin-slab-casting process is the rapid solidification rate necessary to successfully cast this product. In conjunction with the steelmaking modifications discussed previously, rapid solidification results in fine, evenly dispersed globular inclusions. While some refinement of the as-cast austenitic structure was anticipated due to rapid solidification, experience has shown that the grain-size remains large -- on the order of 1 mm -- compared to 2-3 mm for conventional casting. This effect is shown in Figure 1.

DISTRIBUTION OF AUSTENITIC GRAINS

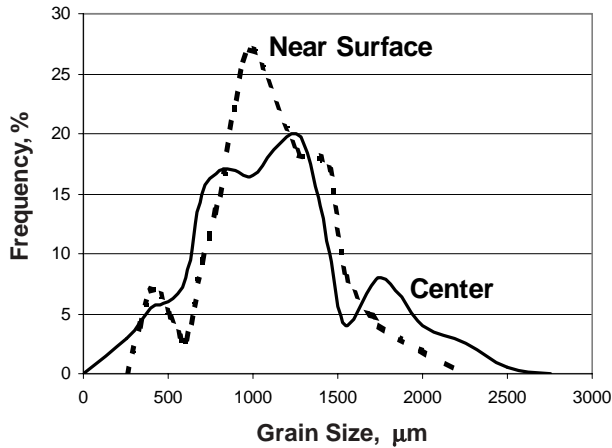


Figure 1 -- As-cast austenite-grain-size distribution in an industrial thin-slab sample¹⁰.

REHEATING AND ROLLING CONSIDERATIONS: GRAIN REFINEMENT AND RECRYSTALLIZATION

Practical design considerations limit the reheating capability of tunnel furnaces used to transfer the slab from the caster to the rolling mill. Operational considerations also dictate that the residence time in the furnace is restricted and subject to variation. Hence, the more appropriate name for the tunnel furnace in a thin-slab process is a holding or transfer furnace, rather than a reheating furnace. The primary operational function of the tunnel furnace is to retain the heat in the slab as it leaves the caster and to equalize the temperature at a consistent level throughout the slab. The maximum reheating-temperature capability for tunnel transfer furnaces is typically between 1100 and 1150°C.

Unlike conventional slab processing, the steel slab is not cooled through a transformation from austenite to ferrite and then reheated back to austenite in the reheating furnace. The benefit of austenite-grain refinement from the γ - α and α - γ transformations during conventional-slab reheating is therefore not available. This necessitates that all austenite-grain refinement and homogenization from the initial dendritic as-cast microstructure must occur during the rolling process. Figure 2 shows the results of a model⁴ predicting the austenite evolution during hot rolling of a 50-mm thick slab of C-Mn steel at 1150°C, near the maximum temperature capability of most thin-slab operations. For a starting austenite grain size of 1000 μm , a 50% reduction is necessary to complete recrystallization in 1 to 2 seconds. At lower temperatures, more time or more deformation is required to complete recrystallization.

Because the initial rolling deformation takes place

EVOLUTION OF AUSTENITE GRAINS

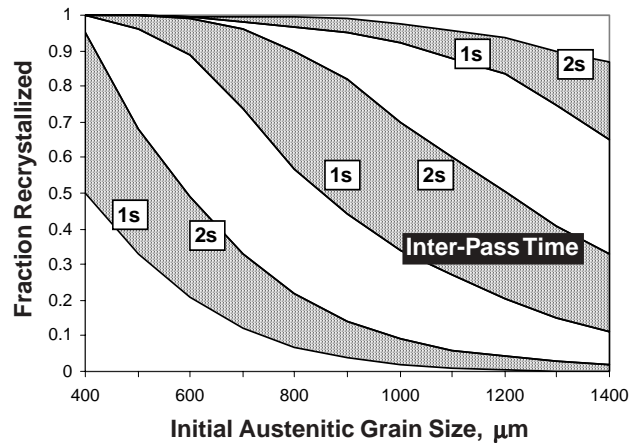


Figure 2-- Austenite evolution during hot rolling of 50-mm C-Mn steel⁴.

with an as-cast microstructure, hot ductility of the slab becomes an issue. Ductility losses from grain-boundary precipitates or grain-boundary penetration of liquid Fe-Cu alloy can promote hot cracking during the initial rolling pass of the dendritic as-cast structure. These grain-boundary ductility issues are aggravated by the large austenite grain size along with natural dendritic segregation during solidification. While higher initial rolling temperatures would be beneficial in minimizing hot-ductility problems during rolling, the limited reheating capabilities of the tunnel furnaces restrict this option.

Figure 3 shows the effect of various microalloying elements on the ductility loss of steels reheated to near melting to simulate the as-cast dendritic microstructure⁵. The key to preventing transverse cracking during casting or edge cracking during rolling is to maintain the temperature during deformation above the point where the ductility loss begins. Figure 4 shows the effect of the reheating temperature on the rate of for-

EFFECT OF TEMPERATURE ON DUCTILITY

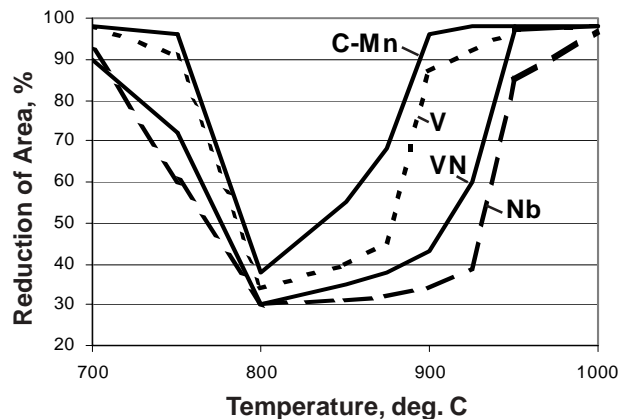


Figure 3 -- Effect of temperature on hot ductility of various microalloyed steels⁵.

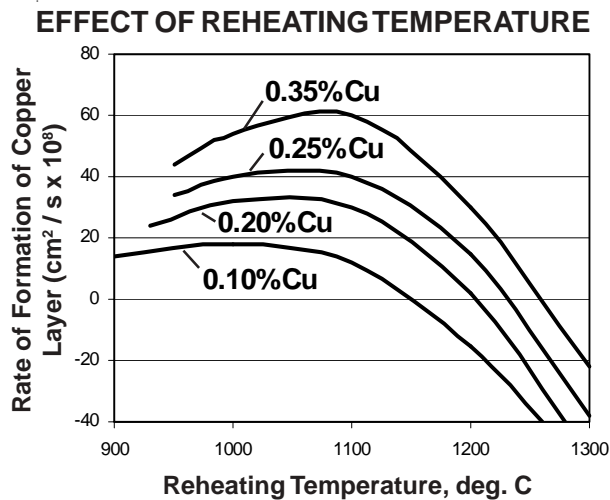


Figure 4 -- Effect of reheating temperature on Fe-Cu layer formation⁵.

mation of the Fe-Cu alloy layer, which contributes to hot shortness. Unfortunately, the temperature for the maximum rate of formation of the grain-boundary penetrating Fe-Cu alloy is the normal operating temperatures of the tunnel-transfer furnaces, 1050 - 1150°C⁵.

For optimum precipitation strengthening, the microalloy should also be in solution in the austenite as long as possible during the rolling process. Maximum precipitation strengthening occurs either in the interphase region during transformation, or even more effectively, in the ferrite after final transformation. Precipitates that form in the austenite are generally larger and more dispersed, reducing the strengthening contribution of the precipitation.

Therefore, the solubility of the microalloy is significant. The microalloy must be in solution in the reheating step in order to contribute to precipitation strengthening in the final as-rolled product. It should remain in solution throughout the rolling process for maximum precipitation strengthening and to minimize deformation resistance caused by precipitation in the austenite that can lead to increased rolling forces, difficulty in controlling shape, and roll wear.

PREFERRED MICROALLOY CHARACTERISTICS FOR THIN-SLAB CASTING

Based on the previous discussion, the preferred characteristics of a microalloy for use in a thin-slab casting process can be summarized as follows:

- Compatibility with EAF steel, particularly with higher nitrogen levels.
- Minimal precipitation during solidification for enhanced castability.
- Highly soluble in austenite at tunnel-furnace temperatures.

- Minimal impedance to static recrystallization during rolling.
- Precipitation for strengthening should occur after finish rolling.
- Promote fine ferrite grains without need for rolling below the T_{nr} temperature.
- Thermal and mechanical processing requirements should not cause undue production complications. (Reduction schedule, finish-rolling temperatures, and coiling temperatures should be similar to carbon steel.)

CHARACTERISTICS OF AVAILABLE MICROALLOYING ELEMENTS

Solubility Considerations

For the desired characteristics for microalloys in the thin-slab process, the solubility in austenite is a critical factor. Figure 5 shows the relative solubility of the carbides and nitrides of V, Nb, Ti, and Al. For the alloys V and Ti, the carbide form is considerably more soluble than the nitride form. For Nb, the difference is less although the nitride is still less soluble at equivalent concentrations of C and N. Vanadium carbide (VC) is dramatically more soluble than any of the other carbides, and vanadium nitride (VN) is more soluble than the other nitrides.

Figure 6 on the next page shows the relative amounts of V, Ti, and Nb combined (precipitated) as an M(C,N) for a range of temperatures for a typical multiple microalloyed steel composition. Over the typical slab temperature in a tunnel furnace (1050°C to 1150°C), the Ti will be fully precipitated, the Nb precipitation will vary widely depending on temperature

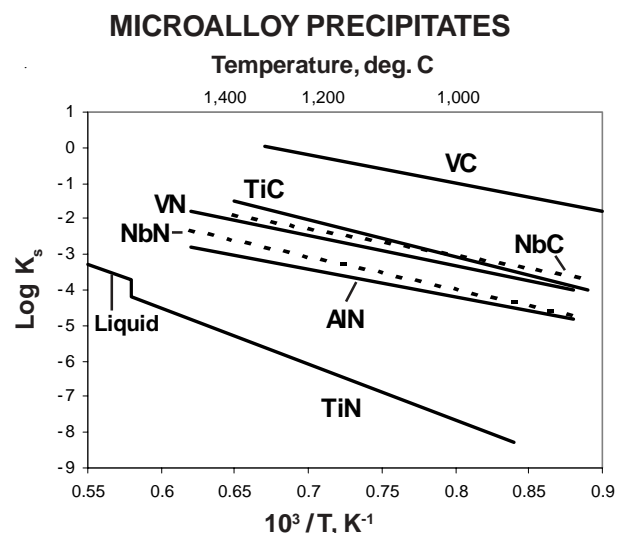


Figure 5 -- Solubility product of microalloy carbides and nitrides¹².

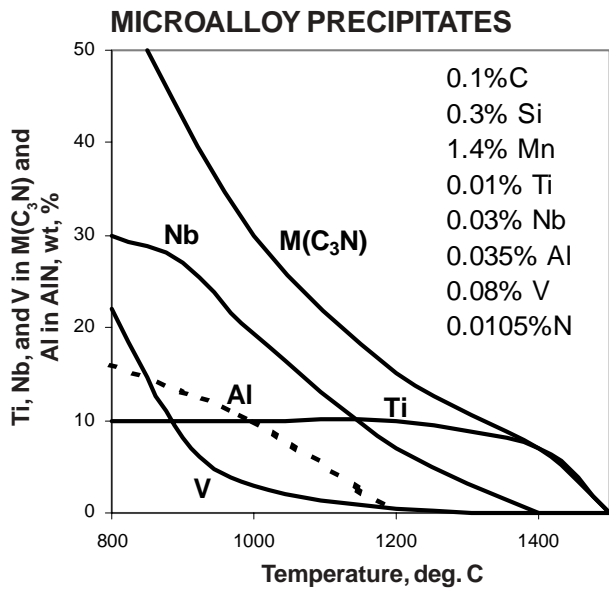


Figure 6 -- Precipitation of carbonitrides in microalloyed steel⁶.

but will never be fully in solution, and the V will almost all be in solution. The preferred condition, as discussed earlier, is that the microalloy be in solution in the austenite at the start of rolling. Only the vanadium will meet the criteria of being in nearly full solution for the conditions typical for rolling thin slabs.

Recrystallization Considerations

Microalloying additions should not significantly impede the necessary static recrystallization needed after the first reduction pass. Static recrystallization can be inhibited by precipitates that are present in the as-cast austenite on entering the first roll stand, or by strain-induced precipitates that form immediately from the first rolling process⁷. As discussed previously, vanadium has the best chance of staying in solution during this initial rolling stage because of the higher solubility of both the nitrides and carbides compared to the alternative microalloys.

Solute drag from the microalloy in solution can also retard static recrystallization of the as-cast structure. Figure 7 shows solute retardation parameters (SRP) for various microalloying elements^{8,9}. These SRP values quantify the delay produced in recrystallization time by the addition of 0.01% of each alloy element in solution with respect to a C-Mn base steel. As shown, vanadium has the least effect of retarding recrystallization, and niobium has the greatest effect. The potential for a microalloying element to delay recrystallization is critical because of the necessity of insuring static recrystallization after the first reduction pass. Either additional reduction and/or higher tem-

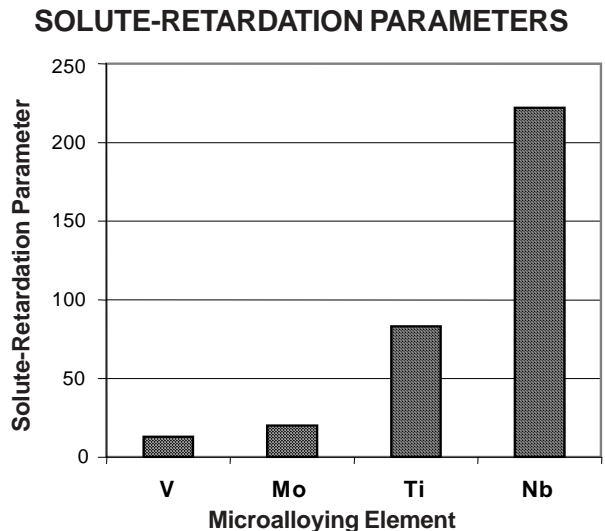


Figure 7-- Solute-retardation parameters for static recrystallization (per 0.01%)^{8,9}.

peratures must offset any recrystallization retardation caused by microalloys.

Uranga¹⁰ noted that decreasing Nb content notably improved the microstructural homogeneity of thin slabs by reducing the fraction of coarse grains and also the size of the largest grains. Smaller reductions in Nb could be tolerated if the entry rolling temperature were increased. Ti has an intermediate SRP value between V and Nb. However, TiN precipitates will exist in the as-cast microstructure and will likely inhibit recrystallization. Vanadium, because of its high solubility and low SRP value, has little effect on the recrystallization process during the initial rolling phase. Thus, the combination of high solubility of the V(C,N) precipitates and the low contribution of V in solution to solute drag make vanadium the desirable microalloying element for minimal retardation of the necessary recrystallization and homogenization of the dendritic as-cast austenitic microstructure.

Recrystallization Considerations

As discussed earlier, the nitrides of V, Ti and Nb tend to form before the carbide precipitates. Vanadium is unique in that nitrogen-rich V(C,N) is the preferred precipitate for effective ferrite strengthening. In the case of Nb, nitrogen will encourage the strain-induced precipitation of nitrogen-rich Nb(C,N) early in the rolling process. Nb is then removed from solution that would otherwise be available for precipitation strengthening of the ferrite⁸. The maximum effective Nb addition level is limited by the maximum reheat temperature available, nitrogen levels, and the ability to achieve sufficient recrystallization of the dendritic as-cast structure. Hensger and Flemming³ reported that V addi-

tions are necessary to achieve strengths above 500 MPa, and maximum strength steels were nitrogen-enhanced to optimize the V(C,N) strengthening in the ferrite.

TiN will tend to form during or soon after solidification and may contribute to strengthening only if the cooling rate is sufficient to keep the precipitates small¹¹. Generally, TiC can be effective for precipitation strengthening, but the higher nitrogen levels of EAF steels make it difficult to keep Ti in solution for subsequent TiC precipitation.

The presence of increased nitrogen in steel reduces the relative strengthening effectiveness of both Ti and Nb additions. The higher nitrogen levels of EAF steels typically used with thin-slab casting become a positive factor when combined with V additions, increasing the effectiveness of the V addition. Strengthening coefficients for N in V steels have been reported to be as high as 7 MPa (1 ksi) for each 10 ppm increment of N as long as the V:N ratio is greater than 4:1¹². Figures 8 and 9 show the effects of nitrogen on strengthening vanadium steels as reported in the literature.

Grain-Size Considerations

For maximum strength and ductility of polygonal ferrite microstructures, it is advantageous to produce as fine a ferrite grain size as possible. To achieve ferrite grain refinement in the final as-rolled product, the austenite should be “conditioned” during the later stages of the rolling process to provide the maximum amount of grain-boundary surface area. One way this conditioning can be accomplished is by finish rolling at temperatures below the recrystallization stop temperature, thereby maintaining the flattened or “pancaked” aus-

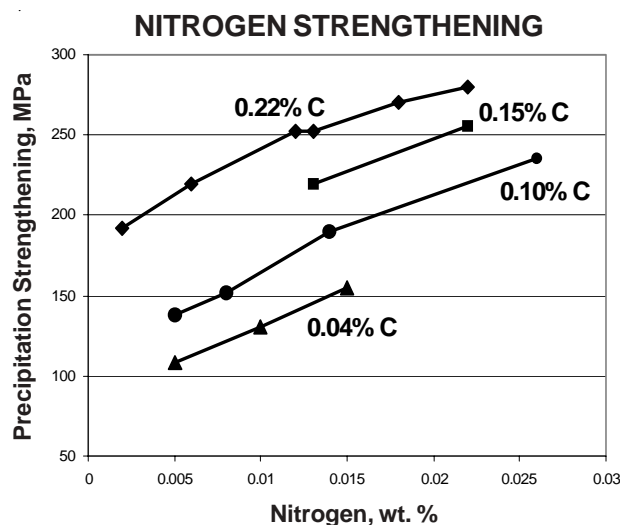


Figure 8-- Precipitation strengthening of 0.12% V steels isothermally transformed at 650 deg. C / 500s¹⁵.

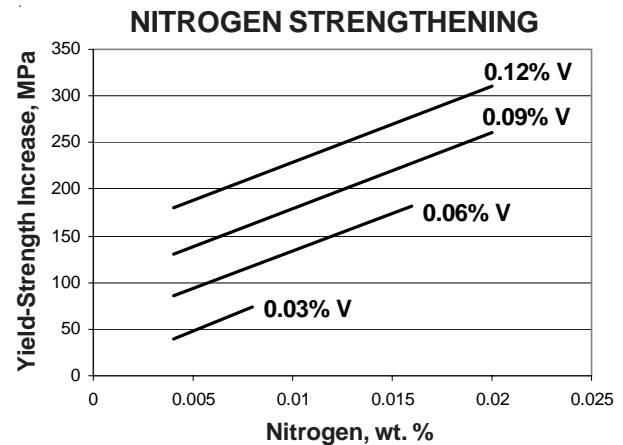


Figure 9-- Effect of N content on yield-strength increase based on model of V-N microalloyed strip¹².

tenite grains that occur from rolling deformation. These flattened grains then promote the formation of small ferrite grains in the final product.

Nb-alloyed steels using these “controlled-rolling” (CR) processes were developed over 30 years ago. In the CR process, the temperature for the final rolling passes must be held below the maximum temperature at which recrystallization will not occur, the “recrystallization stop temperature, T_{nr}”. In thin-slab processing, the use of Nb to raise the T_{nr} temperature is complicated by the previously described problem of Nb also inhibiting the necessary recrystallization of the as-cast microstructure.

Recrystallization controlled rolling (RCR) was developed later. The RCR process relies on repeated interpass recrystallization to achieve a fine austenite grain size, with controlled temperatures and reduction schedules to achieve the desired results. The RCR rolling process involves finish-rolling temperatures more typical for plain-carbon steels, while the CR process requires lower finish rolling temperatures. These higher rolling temperatures promote repeated recrystallization. Since niobium has been shown to raise the recrystallization stop temperature through precipitation and solute drag, V microalloying was naturally more compatible with the RCR process. V does not inhibit the austenite recrystallization required in the RCR process.

Both CR and RCR rolling generate an austenite with a large amount of grain-boundary surface area (austenite interfacial area), promoting transformation to a small ferrite grain size. These small ferrite grains contribute to both strength and to increased toughness. The effect of the ferrite grain size on the final properties of the steel is independent of the austenite conditioning process used to achieve that ferrite grain size. At an equivalent ferrite grain size, the contribution of grain size to the final properties (strength and

toughness) will be comparable. Therefore, the test of the process is the final ferrite grain size and not necessarily the condition of the austenite that produced the ferrite. For example, the presence of V and N can change the ratio of given austenite grain size $D\gamma$ to the transformed ferrite grain size $D\alpha$, as shown in Figure 10. A higher $D\gamma/D\alpha$ ratio means that for a given austenite grain size, the ferrite grain size will be smaller. In other words, the austenite grain size is not the only controlling issue in determining the final ferrite grain size¹³.

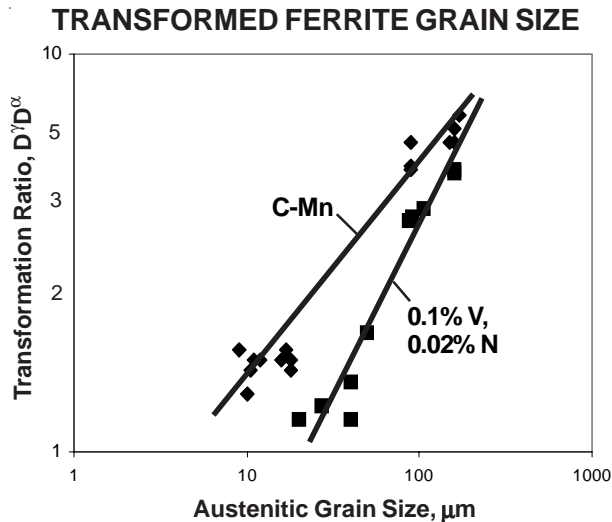


Figure 10 -- Effect of V and N on refining transformed polygonal ferrite size¹⁵.

The evaluation of production HSLA steels from compact-strip processes has shown that an extremely fine ferrite grain size can be produced with any of the normal alloy designs. Figure 11 shows ferrite grain size as a function of gauge for a series of as-rolled V-N

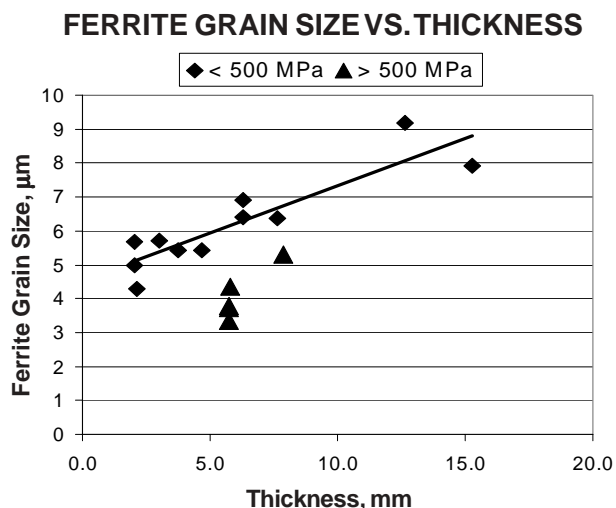


Figure 11 -- Ferrite grain size of production samples of V-N thin-slab strip steels of various yield strengths¹².

microalloyed-strip steels produced on several different thin-slab cast mills¹². The grain sizes for the higher yield-strength heats (>500 MPa) were in the 3 to 5 μm range, which is typical of the best found in any microalloyed-production steels.

Strain-Aging Considerations

Strain aging is often a concern in high-N steels. Strain aging is the increase in flow stress caused by aging a strained material at slightly elevated temperatures. Nitrogen aging can occur from high ambient temperatures upward, while carbon aging generally starts at 150°C. With significant aging after cold deformation, the toughness and ductility of the steel can be substantially degraded. The important consideration here is that it takes “free” or uncombined interstitial carbon or nitrogen to cause this aging phenomenon. Nitrogen that is tied up as a nitride in the steel will not contribute to strain aging.

Strain aging can be evaluated as shown in Figure 12. The strain-aging index, ΔY , is the increase in flow stress after aging a pre-strained sample¹³. The pre-strain was selected as 7.5%, with aging taking place at 100°C to insure that aging was only due to the presence of free nitrogen.

STRAIN-AGING-INDEX TEST PROCEDURE

- ΔY = CHANGE IN YIELD STRESS DUE TO STRAIN AGING
- ΔU = CHANGE IN ULTIMATE TENSILE STRENGTH DUE TO STRAIN AGING
- ΔE = CHANGE IN ELONGATION DUE TO STRAIN AGING

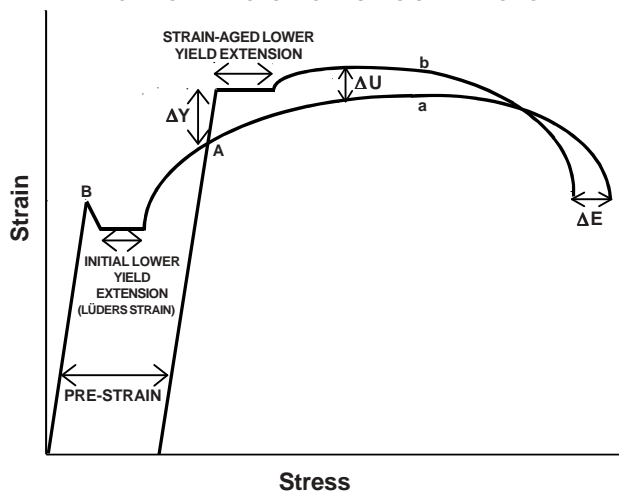


Figure 12 -- A schematic drawing of aging-index (ΔY) test procedure¹⁵.

Figure 13 shows the results of a strain aging evaluation of sheet steels from thin-slab strip mills with EAF melting furnaces. There were from 5 to 10 different heats for each group. The results show that even with average nitrogen values of 150 ppm, there is no strain

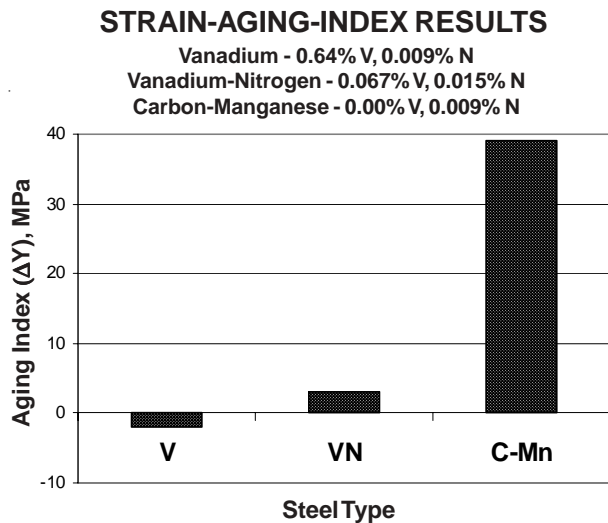


Figure 13 -- Strain aging index results from V, V-N, and C-Mn steels¹⁶.

aging in the vanadium steels. The C-Mn steels, all of which were Al-killed, still had a significant strain-aging response. In as-rolled strip steels, V is much more effective in removing free nitrogen from solution than is Al. Some of the V-N steels contained over 200 ppm nitrogen and still did not exhibit strain aging.

CONCLUSIONS

Based on the previous discussion, vanadium has the capability of providing the stated microalloy characteristics preferred for the thin-slab-cast, direct-charging stripmaking process. Using the previous seven characteristics as a reference, the advantages of vanadium are summarized as follows:

- The higher N levels of the EAF process are used to an advantage to provide efficient precipitation of V(C,N). The nitrogen-rich precipitates of other microalloys are generally ineffective for precipitation strengthening.
- Hot-ductility tests show that V is less susceptible to cracking than Nb steels, even with enhanced nitrogen levels.
- Solubility calculations show that V has the highest solubility of all the microalloying elements, minimizing the need for high-reheating temperatures.
- V has the lowest solute-retardation parameter value, minimizing its effect on retarding the necessary recrystallization of the coarse dendritic as-cast microstructure.
- V stays in solution through the final-rolling pass, minimizing any increase in roll force requirements which can lead to roll-wear and shape-control difficulties.
- Fine ferrite grains down to 3-5 μm are achiev-

able without the need to roll below the recrystallization stop temperature.

- Processing requirements (reheating temperature, rolling temperature, roll pass scheduling, and coiling temperature) are comparable to processing C-Mn steels and do not require major modifications.

With the high degree of competitiveness in the current world market for strip steel, each steelmaking operation will need to find the particular market niche that suits the capabilities of that operation. High levels of production in North America have confirmed that thin-slab casting is particularly suited to producing hot-rolled and microalloyed HSLA steels at competitive costs. The high compatibility of the V and the V-N alloying system to the production process requirements of thin-slab mills, as reviewed in this paper, has made it the alloy of choice for most of these mills.

With over ten years of history and continuous improvement, vanadium-microalloyed HSLA strip steels from thin-slab casters are continuing to expand their usage into even more stringent quality applications. There is no known limit to prevent these mills from supplying the majority of the market for HSLA hot-rolled strip. The cost effectiveness of the V-N microalloying system, along with the cost efficiencies of the thin-slab-casting operation, provide a highly competitive process that will be the standard for the future.

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