THE INFLUENCE OF AUSTENITE ON THE TOUGHNESS OF MARTENSITIC STEELS

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ABSTRACT

The retention or introduction of austenite has often been suggested as a useful way to improve the toughness of high strength martensitic steels. Whether this approach is viable, however, depends on the precise purpose the austenite is intended to serve, and the precise manner in which it is used. This paper discusses the three principal ways in which austenite has been proposed as a toughening agent: suppressing the ductile-brittle transition in lath martensitic steels, improving resistance to hydrogen embrittlement, and increasing the toughness of high strength steels above the ductile-brittle transition. The use of austenite to increase the toughness of high strength steels in the ductile mode is usually illusory; the concomitant loss of strength more than accounts for the gain in toughness. The suppression of hydrogen embrittlement by austenite is a real effect in steels that fail by intergranular fracture along contaminated grain boundaries, but the effect is neutral, and can be negative in the more resistant high strength steels that fracture in a transgranular (interlath) mode. The acknowledged and important role of austenite is in its positive influence on the ductile-brittle transition temperature of lath martensitic steels. But in this case also the austenite effect is subtle and the austenite phase must be precipitated in the right amounts and microstructural locations. Used properly, it lowers T_B by refining the effective grain size of the steel.

INTRODUCTION

For more than fifty years steelmakers have intentionally introduced austenite into high strength martensitic steels to improve their toughness. Many of these efforts have been guided by well-planned schemes for alloy design. Others have been purely empirical, with the metallurgist sometimes learning only after the fact that austenite was even present. However, the decades of research and experiment have served to clarify the role (or, more properly, the roles) of austenite in martensitic steels. The present note summarizes the author's understanding of the present state of the subject.

To achieve a martensitic steel with a good strength-toughness combination one must ordinarily accomplish two things: the steel must fracture in a ductile mode (that is, the ductile-brittle transition temperature, T_B, should be below the intended temperature of service) and the ductile fracture toughness should be high. To optimize the use of austenite, we need to understand its effect on each of these aspects of fracture.

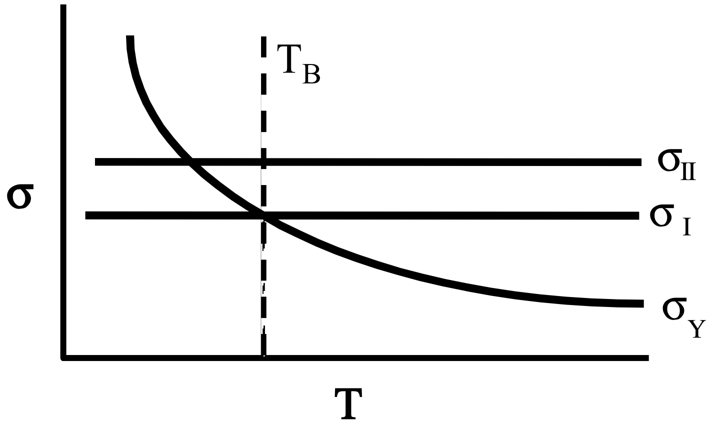

Fig. 1: The Yoffee diagram, illustrating the interplay of yield and fracture stresses in determining T_B.

The ductile-brittle transition is most easily understood from the Yoffee diagram (Fig. 1) [1] which is a bit oversimplified, but provides a very useful framework for understanding the metallurgy of brittle behavior. Lowering the temperature increases the yield strength of steel, and, hence, increases the peak tensile stress in the region of intense stress head of a threatening crack of flaw (which is of the order 3-5σ_Y in the case of a sharp crack in a ductile material [2,3]). Brittle fracture intrudes when the stress rises above that needed to drive brittle fracture. A brittle crack can propagate through bcc iron either by transgranular cleavage or intergranular fracture, and the ductile-brittle transition is controlled by the easier of these.

It follows from the Yoffee diagram that there are two generic ways to lower the ductile-brittle transition. One may lower the yield strength of the steel, or raise the brittle fracture strength. The latter approach is, of course, preferable since it maximizes structural strength. If the easiest brittle mode is intergranular, the reason is usually some chemical contamination of the prior austenite grain boundaries of the martensitic steel. In this case the brittle fracture strength can be raised by purifying the steel or gettering the impurities into relatively innocuous inclusions. A successful result will lower T_B until transgranular cleavage intrudes as the preferred brittle mode. There are then two common ways to increase resistance to transgranular cleavage: refining the prior austenite grain size or introducing a controlled distribution of austenite into the martensite phase [4,5]. As discussed below, the two mechanisms accomplish very similar things.

Conversely, any process that lowers the brittle fracture stress in an alloy of given strength raises T_B. The most common threat is hydrogen, introduced from the environment as a by-product of anodic dissolution (“stress-corrosion cracking”) or cathodic charging (“hydrogen embrittlement”). Austenite has been frequently suggested as a possible way of mitigating this problem.

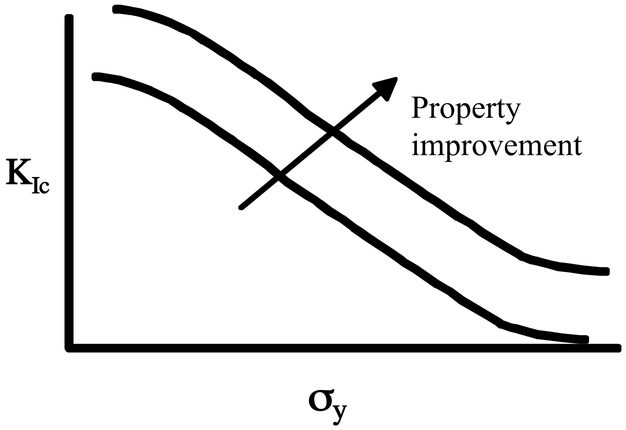

Fig. 2: The strength-toughness characteristic of high strength steel.

Finally, when the steel fractures in a ductile mode above T_B its fracture toughness is a strong function of yield strength (the strength-toughness characteristic illustrated in Fig. 2). It follows that one can always raise the fracture toughness by lowering the strength; the challenge is to raise the strength-toughness characteristic as a whole so that the toughness can be increased without sacrificing strength. It has been suggested periodically that austenite may be useful for this purpose.

AUSTENITE IN MARTENSITIC STEELS

There are two ways to introduce islands of austenite into the dislocated lath martensitic steels that are the best candidates for high-strength service. First, particularly in high-carbon or high-alloy steels a significant austenite fraction is often retained in the form of thin lamellae that separate martensite laths. Second, austenite can be precipitated along martensite lath and packet boundaries by intercritical annealing or tempering in the two-phase ($\alpha+\gamma$) region of the phase diagram.

Retained austenite can be a useful addition to low-strength steels. It transforms to martensite under load, and the associated transformation strain can make a significant contribution to the overall tensile elongation of the steel. Low-strength, high austenite “trip” steels are under active development for their formability in automotive or other manufacturing applications. However, metastable retained austenite transforms under load to a brittle fresh martensite that ordinarily decreases the toughness of high strength steels. As a consequence, high strength steels are ordinarily processed to minimize or eliminate retained austenite. The austenite that is intended to toughen high strength steels is introduced by precipitation during intercritical heat treatments.

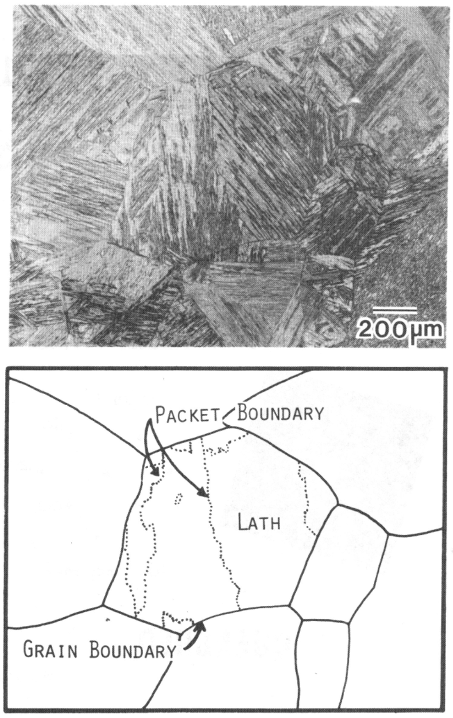

Fig. 3: As-quenched condition of a dislocated lath martensitic steel. Prior austenite grains are divided into packets of parallel laths.

It is useful to divide the intercritical treatments into two categories [6,7]. Both begin from an "as-quenched" (Q) structure like that shown in Fig. 3, in which a prior austenite grain is subdivided into "packets" of parallel dislocated martensite laths that share a common variant (or are themselves subdivided into "blocks" whose laths share a common variant). A block is essentially a single grain of martensite that is made up of laths separated by low-angle boundaries. Intercritical treatments introduce islands of austenite along the prior austenite grain boundaries, packet and lath boundaries.

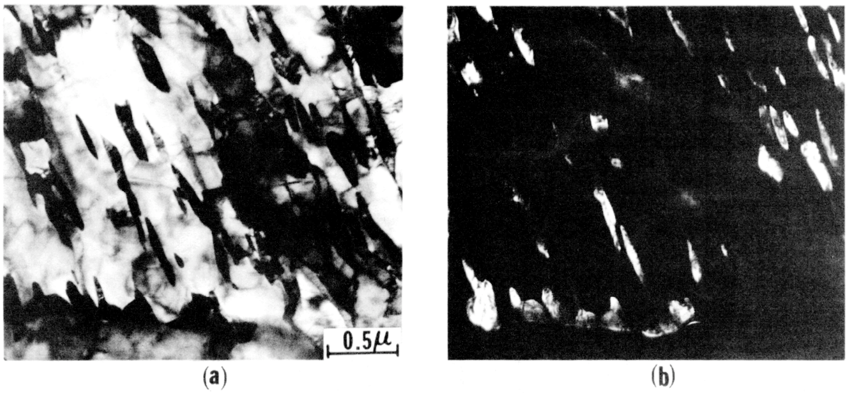

Fig. 4: (a) Electron micrograph of lath martensitic steel after intercritical tempering. (b) Dark-field image; precipitated austenite appears bright.

The distinguishable cases are:

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(1) Intercritical tempering (T). We use this term to denote tempering at a temperature slightly above A_{c1}, in the lower part of the two-phase region. During this treatment a solute-rich γ-phase precipitates while the residual α loses solute and tempers. Most commonly, the γ forms as small islands along the lath boundaries of the martensite. Because of its high solute content and small size, most of the austenite is retained on subsequent cooling to room temperature, producing a microstructure like that shown in Fig. 4. The reaction is

$$\alpha' \rightarrow \alpha_T + \gamma \tag{1}$$

where α' designates fresh martensite and α_T labels martensite that has been tempered during the intercritical treatment. The precipitated austenite serves two functions that affect the brittle transition. It softens the martensite constituent by gettering carbon and other solutes. More importantly, it separates the laths within a packet which, in a somewhat subtle way we shall describe below, reefines the effective grain size of the steel.

(2) Intercritical Annealing (L). We use this term to denote a treatment just below A_{c3}, in the upper part of the two-phase region, and use the symbol, L, to designate it. A high volume fraction of γ-phase forms during an intercritical anneal. The γ-phase precipitates preferentially along martensite lath boundaries, producing a microstructure of parallel laths of alloy-rich γ and well-tempered, alloy-lean α_T. However, the γ-phase formed at temperatures near A_{c3} is only slightly enriched in alloy content and largely retransforms during cooling, producing a "dual-phase" structure that is a mixture of tempered and fresh martensite. The dominant reaction is

$$\alpha' \rightarrow \alpha_T + \gamma \rightarrow \alpha_T + \alpha' \tag{2}$$

There is often with a small amount of retained austenite.

While intercritical annealing produces a dual-phase microstructure it does not significantly refine the grain size. The reason is that the austenite precipitated along a lath boundary has a strong tendency to retransform into the particular variant of martensite that defines the surrounding packet [8]. While the microstructure appears irregular a block or packet in the “dual-phase” structure is essentially a single crystal.

However, the martensite packet is chemically heterogeneous after intercritical annealing. The α_T is relatively lean in solute while the fresh α' is relatively rich [9]. The chemical redistribution significantly affects the response to further heat treatments.

The heat treatments that are actually used to introduce austenite into in high strength martensitic steels are combinations of the treatments described above. They include the "QT" treatment of "9Ni" steel, the "QLT" treatment of low-alloy cryogenic steels and the "LQT" treatment of AerMet 100 [10], an ultra-high strength secondary gardening martensitic steel [11]. We shall briefly describe the essential features of these treatments.

The QT treatment. For over fifty years the "workhorse" ferritic cryogenic steel has been "9Ni" steel, an alloy with nominal composition Fe-9Ni-1Mn-0.06C. It is normally processed through a QT treatment: austenitize at 800 ºC for 1 hr., quench (Q), then temper at 570-600 ºC for 1 hr. (T). The Q treatment produces a dislocated lath martensite with a reasonably small effective grain size. The T treatment tempers the α' and precipitates 5-10% γ, primarily along the lath boundaries. The regular pattern of interlath austenite is illustrated in Fig. 4. In this condition, 9Ni steel has a ductile-brittle transition temperature below 77K. As we shall discuss below, the intimate intermixture of tempered martensite and interlath γ is largely responsible for this behavior.

The QLT treatment. Given that Ni is the costly alloy addition to 9Ni steel, there is an obvious economic incentive to reduce the nickel content. For this reason cryogenic steels of (5-6)Ni content were developed and steels with Ni contents as low as 3.5Ni have been investigated [6,10]. These steels are processed with variants of the QLT treatment. The most thoroughly researched is the Nippon 5.5Ni (typical composition: Fe-5.9Ni-1.2Mn-0.7Cr-0.2Mo-0.06C), which can serve as an example [12].

The objective of the QLT treatment is to reproduce the microstructure of 9Ni steel in an alloy of lower Ni content. In particular, the microstructures should replicate the dense distribution of stable austenite precipitates along the lath boundaries. However, if a 6Ni steel is given a 1-2 hr. intercritical temper to mimic the treatment of 9Ni steel very little austenite is formed, and that which does appear is lean in solute and thermally unstable. The fraction of austenite can be raised by increasing the tempering time; if Nippon 5.5Ni is tempered for 100 hrs. at 600°C it develops about 8 volume percent of stable precipitated austenite. But this austenite is blocky and dispersed in the microstructure and does not effectively disrupt the crystallographic alignment within the martensite packets. Both short and lengthy tempers leave the ductile-brittle transition well above 77K.

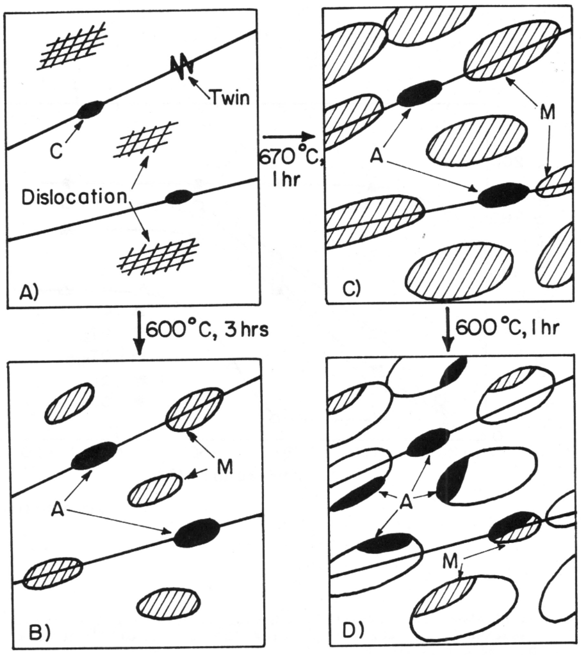

Fig. 5: Schematic of the microstructural evolution during a QT (a-b) and QLT (a-c-d) treatment of a 6Ni steel [12].

The solution to this problem was the QLT heat treatment, which interposes an intercritical anneal (L) before the precipitation temper. The microstructural changes that occur during the QLT treatment are diagrammed in Fig. 10. During the L step the alloy decomposes to a mixture of austenite and tempered martensite. Because of the high temperature of the L treatment, most of the austenite reverts to martensite on cooling. While T_B remains high, the austenite formed during the L treatment is high in Ni [27]; the resulting microstructure is a "dual-phase" structure of low-alloy laths (the original martensite) and high-alloy laths (the fresh martensite). During the intercritical temper that completes the QLT treatment austenite nucleates readily along the boundaries of the high-alloy laths, which behave essentially as if they were 9Ni steel. The result is a dense distribution of stable precipitated austenite. Since the high-alloy lath boundaries are dense within the martensite packets, this austenite effectively decomposes the packets and substantially lowers T_B

Both the microstructure and the strength-toughness combination of QLT 5.5Ni at 77K are similar to those of 9Ni steel in the QT condition.

A similar approach was used in the design of 3.5Ni steel. However, in this case the lower Ni content makes it more difficult to assemble high-Ni laths that will form stable austenite on tempering. The successful heat treatment can be designated QLL'T, and involves two successive intercritical anneals, the first at a relatively high temperature to form laths of intermediate Ni content and the second at a lower temperature to form high-Ni laths within these. The final temper causes the precipitation of a dense, stable precipitated austenite.

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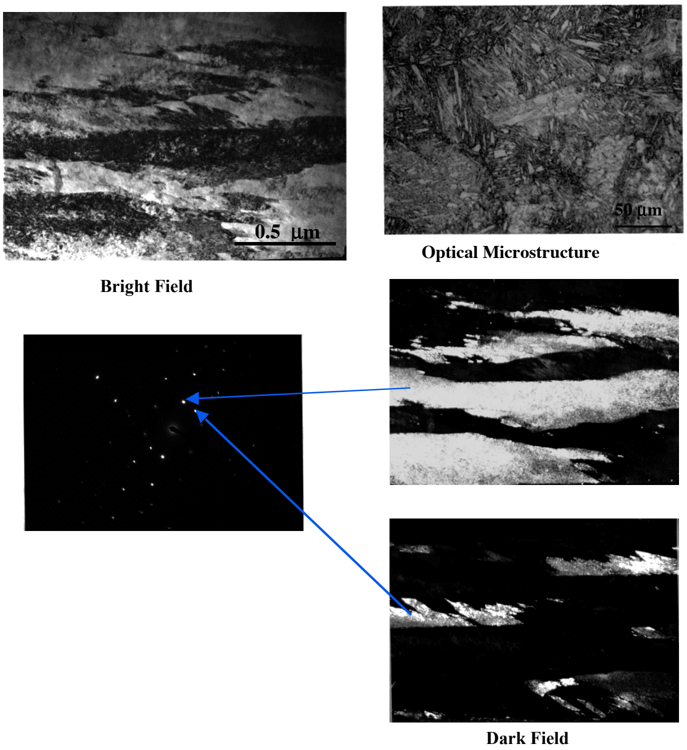
The LQT treatment. While the QLT treatment breaks up the martensite packet by introducing precipitated austenite, the reversed cycle, LQT, refines the packet structure directly before precipitating stable interlath austenite. It is used, for example, in some heats of AerMet 100 to create a fine grain size that ensures ductile fracture at very high strength levels [10].

To understand how intercritical annealing can lead to grain refinement, let an L-treated steel be given a reversion (Q) treatment. The result is illustrated in Fig. 6, which shows AerMet 100 in the "NLQT" condition, where "N" denotes a long anneal at 1150 ºC that creates a homogeneous, but very coarse initial microstructure.



Bright Field

Optical Microstructure

Dark Field

Fig. 6: Microstructure of AerMet 100 in the "NLQT" condition showing multiple variants of martensite.

On heating into the austenite field, both constituents of the L-treated steel revert to austenite. If the alloy is held in the γ-field for some time after reversion, as it must be if the part is a thick plate, the carbon may redistribute. However, the low diffusivity of the substitutional species in the γ-phase prevents their homogenization. The "dual-phase" character of the alloy is preserved. On subsequent cooling, the "dual-phase" alloy undergoes a two-step martensitic transformation. The low-alloy constituent transforms first. Since it is constrained by the surrounding austenite during its transformation, it is severely worked, and deforms the austenite in turn. The high-alloy phase then transforms, while constrained and deformed by the martensite that has already formed. The result is that the transformation occurs under severe mechanical constraint, which encourages local volumes to transform into the martensite variants that are most compatible with the local stress rather than those that continue the pattern in a martensite packet. The result is a very fine grained microstructure.

The microstructure of AerMet 100 processed through an "NLQT" sequence is shown in Fig. 6. The individual laths within the packets are in a fine mixture of two or more martensite variants, yielding an effective grain size of less than 0.5 µm on the {100} planes.

INFLUENCE OF PRECIPITATED AUSTENITE ON THE DUCTILE BRITTLE TRANSITION

The essential role of precipitated austenite in suppressing the ductile-brittle transition is evident from its behavior in 9Ni steel [1,3,7]. In the absence of precipitated austenite the T_B of 9Ni steel is near room temperature. The reason is its effective grain size is coarse. The dislocated lath microstructure in quenched (5-9)Ni steel appears fine-grained in optical or bright-field transmission

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electron micrographs, but is not. The laths within a given packet (or, in some alloys, in “blocks” within packets) are in close crystallographic alignment [5,7,13]. The packet is, essentially, a single crystal of martensite. Its size defines the coherence length in the {100} planes and, hence, the effective grain size. In cleavage, cracks cut across the laths in a packet on their common {100} planes, branching at packet boundaries where the crystallographic orientation changes [4,7].

The mechanism by which precipitated austenite lowers T_B has been a matter of discussion. Possibly the earliest suggestion was the "blunting" effect of the relatively ductile austenite phase. This mechanism was ruled out by the demonstration [8] that the precipitated austenite transforms to martensite ahead of the crack tip, and is not available to blunt a fracture. Moreover, a study of the mechanical properties of an austenitic alloy made to match the composition of the precipitated austenite [14] showed that this material is not ductile at all, but brittle. A second suggested mechanism was "transformation toughening" by the martensitic transformation of the austenite. This possibility is ruled out by the fact that the decrease in T_B is essentially independent of the austenite fraction, and is most pronounced when the precipitated austenite fraction is relatively small.

The mechanism that appears to govern the austenite effect in 9Ni and related cryogenic steels is based on the crystallography of the re-transformed austenite. While the thermal transformation of precipitated austenite in intercritically tempered steel creates martensite that continues the crystallographic variant of the surrounding packet, the variant that forms on transformation under load is ordinarily quite different. For example, let a cleavage crack propagate across a martensite lath on a (001) plane, as illustrated in Fig. 7, and impinge on a layer of interlath austenite that is congruent with it and crystallographically related to it by the Kurdjumov-Sachs relation. It can be shown [15] that the elastic energy increases significantly if the interlath austenite transforms to the crystallographic variant of the parent, but decreases significantly if it transforms to a crystallographic variant that is a different Bain variant of the martensitic transformation [5], that is, a variant that has a different axis of compression of the parent austenite. Since martensite laths that are different Bain variants of the transformation from the same parent austenite do not share {100} cleavage planes, the transformation of precipitated austenite under load dramatically refines the effective grain size

It follows that the transformation of precipitated austenite under load dramatically refines the effective grain size of the martensite packet by introducing lamellae that have a very different crystallography from that of the surrounding packet and do not share {100} cleavage planes with it. The result is to inhibit cleavage fracture and lower the ductile-brittle transition temperature.

This reasoning applies to steels processed through the QT or QLT treatements or any simple variant of them. When the LQT (or some variant of it) is used, there is a similar grain refinement, but in this case the precipitated austenite that is critical to grain refinement is that which is generated during the intercritical (L) teatment of the process.

INFLUENCE OF PRECIPITATED AUSTENITE ON HYDROGEN EMBRITTLEMENT

The influence of precipitated austenite on the hydrogen embrittlement of high strength martensitic steels is superficially unclear, as different researchers have reported dramatic improvements [15,16] and significant deterioration [17] in hydrogen resistance. To understand these conflicting results it is necessary to return to the fracture mechanism and recognize that the hydrogen embrittlement of high strength steels may result in either of two very different failure modes. High strength steels of ordinary purity usually contain grain boundary contaminants at a level that is sufficient to trigger brittle intergranular fracture in the presence of hydrogen [19], so the common mode of failure is intergranular fracture.

When high-strength martensitic steels are purified or gettered, however, the fracture mode in hydrogen shifts to a pseudo-transgranular mode with feathery, brittle facets that traverse the prior austenite grains. On close examination these facets are found to be lath boundary fractures rather than cleavage fractures, and generally follow {110} lath boundary planes rather than {100} cleavage planes [20]. In this case the interlath precipitated austenite that has such a beneficial

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effect on the ductile-brittle transition is actually harmful; the introduction of interlath austenite decreases toughness in the presence of hydrogen [17].

The reason that precipitated austenite suppresses intergranular fracture in steels of ordinary purity seems straightforward. Austenite precipitates along prior austenite grain boundaries and, because of its greater solubility for most harmful impurities, acts to getter these so they are removed from the boundary. The harmful effect of interlath austenite is more subtle. The mechanistic research [17] was done on 5.5Ni (6Ni) steel. In the QT condition, with essentially no austenite, this alloy has a yield strength of 740 MPa and a fracture toughness of 330 MPa√m at ambient temperature. In the QLT condition the lath boundaries of the alloy are densely decorated with precipitated austenite. The yield strength is about 590 MPa and the fracture toughness is about 360 MPa√m. More importantly for the purposes for which that alloy was designed, its cleavage resistance is very high and its ductile-brittle transition is below 77K.

When the alloy is charged with hydrogen, it embrittles in a transgranular mode. Its yield strength increases slightly, but its fracture toughness drops significantly, to 180 MPa√m for the QT condition, and to 100 MPa√m in the QLT condition. The significantly lower toughness of the QLT alloy shows that embrittlement is significantly worse when austenite is precipitated along the lath boundaries. The apparent reason for this behavior is suggested by the profile fractograph shown in Fig. 7.

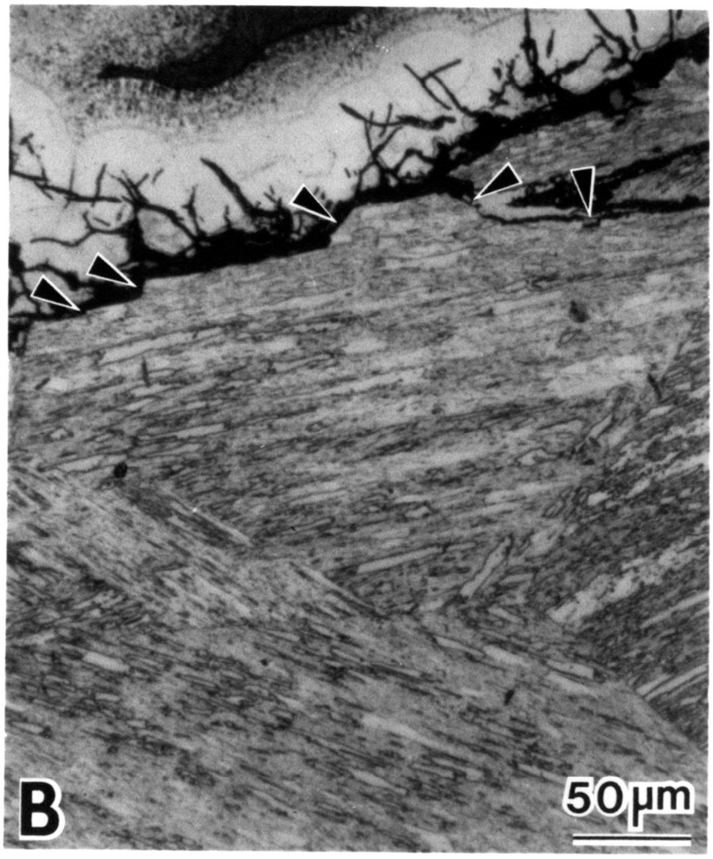

Fig. 7: A profile scanning electron micrograph of a hydrogen-induced fracture in QLT-treated 6Ni steel. The fracture is by lath decohesion along boundaries decorated by precipitated austenite as in Fig. 4.

As illustrated in Fig. 7, the fracture mode in the alloy is interlath. In addition, there are frequent secondary cracks in the lath boundaries just beneath the fracture surface. These invariably emanate from islands of fresh martensite, which are the transformation product of the precipitated austenite formed during the QLT treatment. It appears that the transformation of the austenite contributes to the embrittlement of the lath boundary.

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The probable mechanism by which the martensite transformation contributes to embrittlement is the "wedging" effect of the transformation strain. The martensitic transformation involves a positive volume change, with an associated shear. The transformation of an island of austenite along the lath boundary hence produces a mechanical "wedge" that acts to split the boundary open. If the boundary is already embrittled, this new mechanical load can produce a brittle, interlath fracture of the sort seen in Fig. 7.

INFLUENCE OF PRECIPITATED AUSTENITE ON TOUGHNESS IN THE DUCTILE MODE

One must be cautious in evaluating data suggesting that austenite improves the fracture toughness of high strength martensitic steels in the ductile mode. Much of the available data is deceptive for one of two reasons.

First, retained or precipitated austenite in high strength martensitic steel ordinarily transforms to martensite during mechanical deformation. The associated transformation-induced plasticity increases tensile elongation. This effect can lead to a significant increase in the "toughness" that is estimated from the tension test (for example, the area under the stress-strain curve). However, experience has shown that tensile elongation and tensile "toughness" are not reliable measures of fracture toughness, particularly in the case of metastable austenitic steels.

Second, retained or precipitated austenite almost always lowers the strength of martensitic steel. Since fracture toughness is inversely related to strength via the strength-toughness characteristic (Fig. 2), the fracture toughness usually goes up. The issue is whether the increase in fracture toughness is greater or less than would be achieved by ordinary tempering or other conventional ways of lowering the strength. That is, can austenite be used to improve the strength-toughness characteristic of high strength martensitic steels?

All evidence known to the author supports the opposite conclusion: that retained or precipitated austenite decreases the fracture toughness in the ductile mode. A specific example is contained in recent research on the influence of austenite on the toughness of AerMet 100, and ultra-high strength martensitic steel [21]. In its normal treatment, this steel is given a cryogenic treatment after quenching to minimize retained austenite. Fracture toughness drops dramatically if this treatment is not done. Moreover, austenite can be precipitated in the steel by choosing the tempering temperature. Increasing the fraction of austenite in this way lowers strength and may increase toughness, but leads to lower toughness than is achieved at the same strength level by controlling the degree of precipitation hardening. The cause of the lower toughness is not entirely clear, but may be simply that austenite transformation creates fresh martensite, which acts as a brittle inclusion and source of stress in the high-stress field ahead of a propagating crack.

CONCLUSION

The influence of austenite on the toughness of martensitic steels depends on the particular fracture mode that is of concern. Austenite can be very beneficial in suppressing the ductile-brittle transition provided that it is densely distributed in interlath lamellae that refine the effective grain size when they transform under load. Precipitated austenite is beneficial to the hydrogen resistance of conventional steels, apparently by its action in gettering impurities from grain boundaries. Conversely, it may decrease the hydrogen resistance of very clean steels since its transformation under load promoted interlath decohesion. Retained or precipitated austenite has a generally negative effect on the toughness of high strength steels in the ductile mode. While it may increase toughness by lowering strength, the benefit is less than would be obtained if the steel were soften by more conventional means.

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