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CRACKING OF RAILROAD WHEELS

by

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ABSTRACT

This project was undertaken to summarize known work on the sudden fracture of eutectoid-steel railroad wheels, with the objective of finding new areas for future research into this serious problem. Cracking occurs due to a combination of high thermal stresses, primarily generated during braking, and extreme metallurgical embrittlement. This report focuses on the metallurgical embrittlement aspects of the problem, as stress generation has been studied previously. Since the mechanism of wheel fracture is still not fully understood, this report discusses related cracking problems in steel that seem to occur via a similar mechanism. The remainder of the report explores a wide range of embrittlement mechanisms, ranging from intergranular to transgranular brittle fracture and, finally a summary of fracture toughness studies on the affected eutectoid steel grades. Although numerous possibilities deserving future attention are exposed, the most promising avenues for future research appear to lie in: 1) the development of new alloys that achieve improved fracture toughness through lower carbon content and 2) renewed attention on stress generation.

ACKNOWLEDGEMENTS

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INTRODUCTION

Steel railroad wheels are subjected to a difficult service life since they must both support the mechanical loads from the train car weight and withstand the thermal loads arising from friction between the wheel rim and brake shoe during braking. Since the wheel is a single-load-path component, its fracture may have catastrophic consequences such as derailment of the train. It is therefore very important to determine when a wheel is unsafe or has been subjected to unsafe conditions so that it can be removed from service. In order to do this, it is essential to have a complete quantitative understanding of the mechanism(s) that can lead to catastrophic failure.

Many previous studies have been devoted to understanding wheel failures. Unfortunately, there appears to be several different mechanisms responsible and the most important ones are perhaps the least understood.

In order to obtain high strength and good wear resistance, railroad wheels are manufactured from essentially brittle material--plain-carbon, eutectoid steel. At the same time, the wheel is subjected to loads arising from several sources which can produce significant stress levels. (1) The most important of these are the residual tensile stresses in the rim that can arise after severe braking cycles. This happens when material in the outer layer of the rim yields in compression after expanding during brake heating and subsequently is constrained when it attempts to contract during cooling back to ambient temperature. (2) These residual tensile stresses are then superimposed on the residual stresses produced during manufacturing which may be either tensile or compressive. (3) At the same time, the mechanical loads from the weight of the train produce significant stresses near the contact region between the wheel and the rail. (4) During cornering, additional horizontal loads are imposed on the flange.

Each of these four stress sources are cyclic or periodic, in nature, ranging from the infrequent but high stress-level cycles of repeated brakings to the high-cycle low-amplitude stresses from mechanical loading during each wheel revolution. The stresses arising during the less frequent cornering and braking cycles have an additional cyclic aspect owing to revolution of the wheel during these events.

Considering the combined effects of cyclic mechanical stresses, repeated thermal stresses arising during braking, and a brittle, notch-sensitive steel, it is not surprising that wheels are susceptible to several different kinds of cracking problems.

The various defects exhibited by railroad wheels include:

1. Shelling. Cracks forming just below the tread surface can grow in a circumferential direction, due to the action of contact stresses between the rail and wheel, and eventually causes pieces of the tread to spall out.
2. Spalling. Shallow surface cracks called thermal checks can arise due to martensite formation during severe stop braking cycles which may also lead to spalling or shelling.
3. Wheel Deformation. Relaxation of the prestressed fitting of the cold hub by thermal stresses elsewhere in the wheel can produce shape change that results in loose fits of the hub around the axle.
4. Plate Cracks. Tensile stresses during brake heating cycles in combination with mechanical loads from the axle produce thermal fatigue cracks in the plate, particularly at the front plate-hub fillet, that destroy the load carrying capability of the wheel.

5. Rim Cracks. Thermal cracks or thermal fatigue cracks in the wheel rim can propagate to a significant depth, reduce the service life of the wheel and may initiate catastrophic wheel fracture.

6. Wheel Fracture. Rapid extension of a crack completely through the wheel by brittle cleavage can fracture the wheel into several pieces and may cause a train derailment.

Of these defects, complete wheel fracture is by far the most serious problem. About 100 train derailments occur per year that are attributed to wheel fracture. Not only does an accident of this nature cost an average of $30,000 per incident but the associated potential loss of life and property gives this problem the highest priority. Unfortunately, the causes of this problem are, perhaps, the least understood. The present investigation has been undertaken to review what is currently known about this problem of complete wheel fracture and to explore potentially related literature from other disciplines with the objective of improving the understanding of the metallurgical embrittlement mechanisms that cause these fractures.

PROBLEM DESCRIPTION

The problem of catastrophic fracture of railroad wheels has received a good deal of attention in previous studies. A picture of a fractured wheel is shown in Fig. 1.[2] Based on these previous studies, a fairly clear qualitative understanding of the steps involved in wheel fracture has evolved. Fracture occurs through three distinct stages of crack development:

I. Initiation in the rim of a short, shallow surface crack or cracks running transversely to the tread or flange.

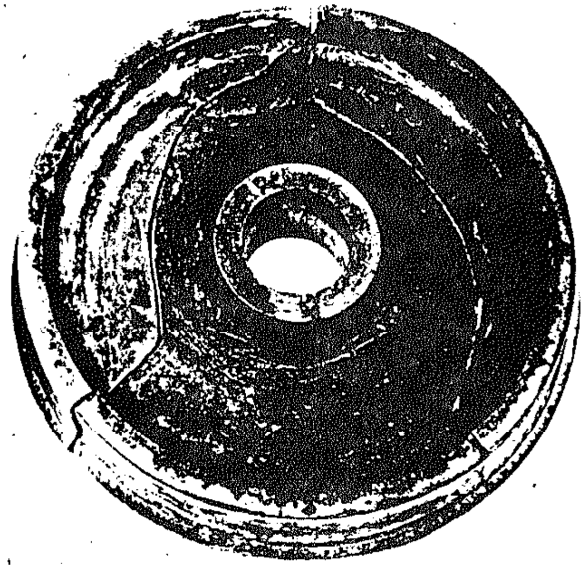
II. Propagation of one or more of these cracks radially inwards through a combination of thermal fatigue and/or cleavage jumps.

III. Overload fracture of the entire remainder of the rim once a crack has reached a critical length.

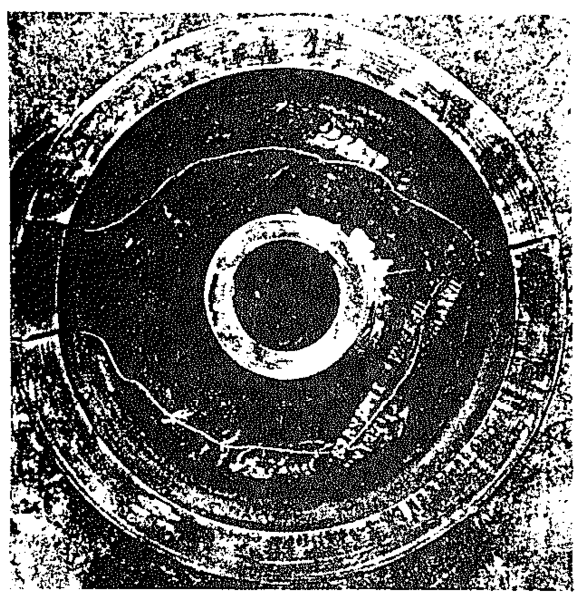
Following a brief metallographic description of the failure, each of these three stages in wheel fracture will be examined in depth, including their most plausible mechanisms.

Metallographic Analysis

Catastrophic wheel fracture has been experienced in all types of wheel--on locomotives, passenger cars and freight cars. It has happened to both cast and wrought steel wheels of all classes, manufactured by a broad range of processing sequences including untreated, rim quenched, whole wheel quenched and heat treated. The wheels are made of plain carbon, eutectoid steel containing .6 to .75% C, .6 to 1% Mn and about .15% Si. They are generally fully pearlitic and contain occasional globular oxide inclusions.



(a) Wheel fracture produced in service



(b) Wheel fracture produced in the laboratory

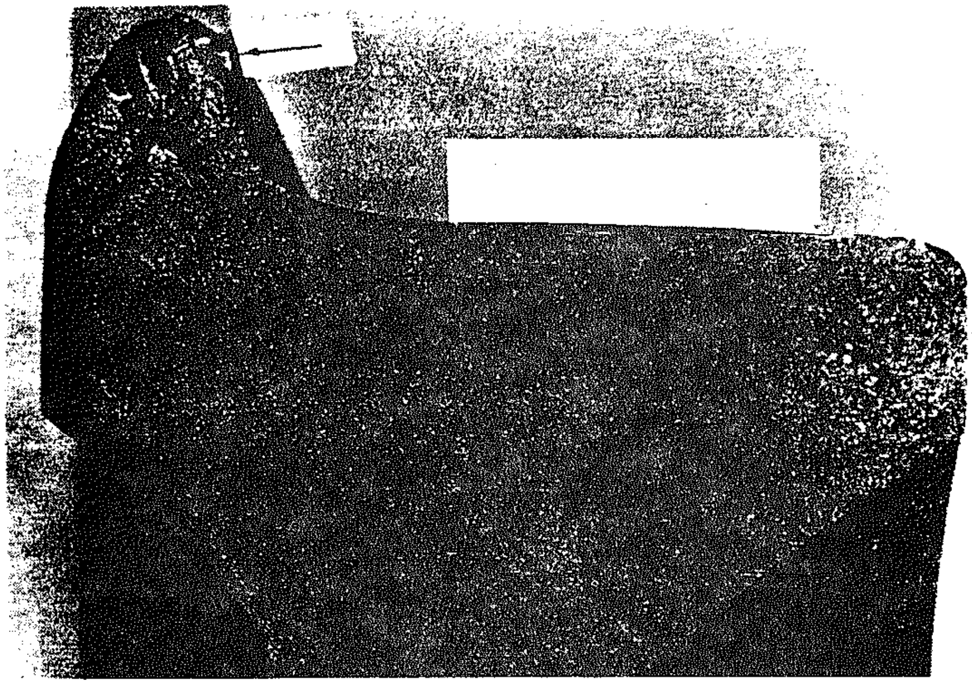
Figure 1 Typical Wheel Fractures

Pictures of the fracture surfaces of two broken wheels shown in Fig. 2 clearly illustrate the three steps of crack formation. The initiation sites are very small and can be found almost anywhere on the rim surface. For the two samples pictured in Fig. 2, arrows point to the initiation sites on the flange and far edge of the tread surface respectively.

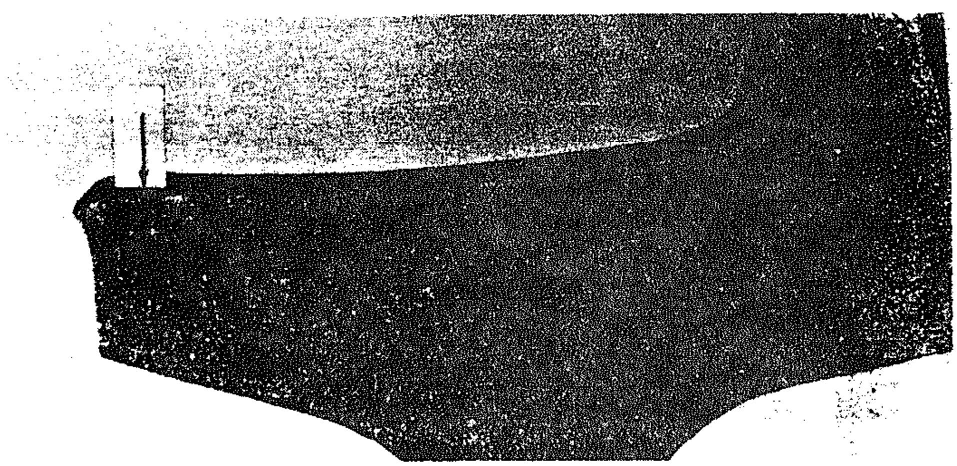
The propagation stage is manifested differently in the two examples. Figure 2a exhibits beech marks progressing through the flange from the initiation site that are indicative of growth by a fatigue mechanism. Figure 2b has a small, semi-circular cleavage zone about the initiation site indicative of an abrupt, brittle, cleavage extension. The cleavage zone forming during the crack propagation stage is often easily distinguished from the overload cleavage zone because it has been darkened by the formation of oxides during subsequent brake heating cycles. Other fractured wheels often exhibit a combination of these two crack models during this propagation stage, with alternating zones of thermal fatigue and cleavage.

The remaining surface of both wheels consists of 100% transgranular cleavage farets that are characteristic of sudden brittle fracture by overload. The lack of oxidation of the overload cleavage fracture surface reflects the observation that final fracture occurs at lower temperatures.

No unusual change in prior austenite grain size, pearlite colony size, inclusions, or segregation has been found associated with the fracture region. The largest cleavage facts and pearlite colony sizes were found in wrought U class wheels. An excellent correlation was found between cleavage facet size and pearlite colony size, but no correlation was found with prior austenite grain size. A final observation is that



(a) Thermal crack growth by a fatigue mechanism in a class-U wheel



(b) Thermal crack growth by a cleavage mechanism in a class-U wheel

Figure 2 Picture of Thermal Cracks - Boeing cracking susceptibility increases with increasing carbon content. The addition of grain refining elements, particularly V, has been reported to improve cracking resistance.[1]

Stage I - Initiation

Catastrophic wheel fractures always appear to initiate from an existing surface flaw. In service, they have been found to initiate at the flange (corner cracks) and along the tread with roughly equal frequency.[1] Somewhat less common are crack initiation sites on the rear rim,[1] often associated with the retarder contact region.[3] A common initiation site is the highly stressed portion of the tread surface at the junction between the flange and tread. Schematic drawings of the different crack types that result are given in Fig. 3.

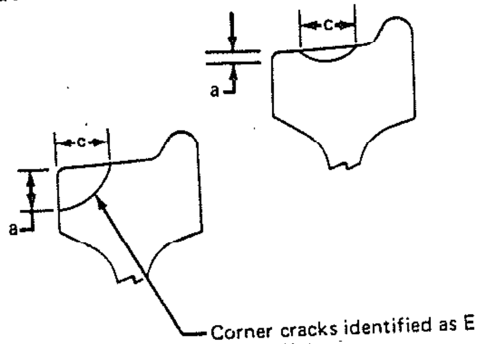
Due to the very severe environment imposed on the tread surface, several different mechanisms are available for initiating cracks.

One of these is thermal cycling of a "hot spot" during the successive wheel revolutions within a single brake cycle.[4] Hot spots begin during a brake application on local portions of the tread surface where a small spot randomly receives more heat input than the surrounding area. This causes proportionately more expansion at this spot, forcing to bulge slightly away from the tread surface. The "hill" that is so created has better contact with the brake so it heats even more, sustaining the same hot spot over many wheel resolutions during a single braking cycle. Crack initiation is often associated with these hot spots on the tread surface.

A second mechanism for initiating shallow cracks in the rim is called thermal checking, and involves a two step process over successive brake heating cycles. During a severe, stop braking cycle, the outer surface

<u>Thermal Crack Types</u>

Tread:



$\frac{a}{c}$ = .5-1 usually

surface cracks

Critical crack length 1-3" (smallest.3" diesel loco class B)

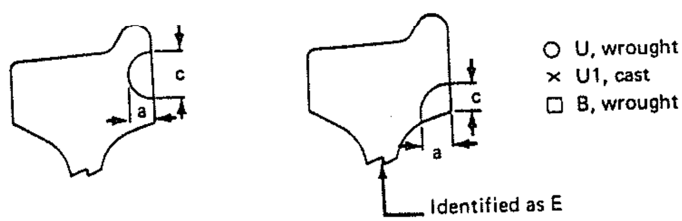
Flange:

$\frac{a}{c}$ = .5-1

Critical crack length 0.5-1" (smallest .2")

Rear Rim:

$\frac{a}{c} \approx 1$



Critical crack length 1-2" (min. 8")

Figure 3 Thermal Crack Types

may heat above the critical A_{c_1} temperature, in localized areas (hot spots) thereby forming austenite. The subsequent rapid quenching by the remainder of the solid mass of the rim after braking is complete may be fast enough to form martensite to a depth of up to 0.1" below the tread surface. The volume expansion associated with martensite formation will force this region into compression, accompanied by plastic yielding. Subsequent heating cycles may temper this martensite, and the accompanying shrinkage can produce tensile stress upon cooling. This, in turn can initiate shallow cracks through the brittle martensite that are known as thermal checks. Although they generally do not propagate to severe lengths, they often cause spalling and some researchers believe they could act as initiation sites for the thermal cracks which cause catastrophic wheel fracture.

Numerous other mechanisms exist for producing crack initiation sites on the rim surface. These include thermal fatigue, possibly combined with rail contact stresses, interaction with the brake pads and stress concentration at machining defects. Since the brake shoes continuously wear away the outer layers of the tread surface, initiated cracks are not a problem unless they propagate deeper into the rim.

Stage II - Propagation

The propagation of initiated cracks is the most critical and least understood phase of wheel fracture. Propagation can occur in two general ways, as stated earlier:

1. Fatigue, whereby the thermal cycles produce "thermal fatigue cracks" exhibiting characteristic beech marks

2. Cleavage, whereby the crack propagates intermittently in rapid transgranular cleavage bursts, or "thermal cracks," interspersed with periods of complete crack arrests.

Cracking usually consists of a series of transverse cracks in the tread that spread radially inwards at different rates. Purely fatigue cracks are more common in the flange and appear to be related to the combined action of thermal stress and the horizontal cornering forces on the flange. Purely thermal cracks are most common in dynamometer tests where cleavage bursts are induced solely through thermal stresses. Cracks in service often propagate through a combination of both thermal cleavage cracks and fatigue cracks. This implies that wheel loads can interact with thermal loads in producing cracks through alternating propagation mechanisms.

The factors that govern the occurrence of cleavage bursts and their arrest are not well understood (relative to fatigue). Clearly, a combination of severe metallurgical embrittlement and high tensile stresses is necessary. However, it is not known which of these is the controlling factor. Possibly, the crack propagates until it meets an unfavorable grain orientation or tougher material. Alternatively, the crack might propagate radially through the tread to an interior layer where the tensile stresses are lower or even zero, at the neutral axis dividing compressive and tensile regions of the wheel. Curiously, the semi-elliptical shape of these cracks corresponds closely to the thermal isotherms arising during brake heating. As the cracks grow, they often increase in depth faster than in width, thereby changing their aspect ratio from semi-elliptical to semi-circular. This is particularly common for cracks initiating on the back rim.[3] This phase of crack propagation is commonly associated with severe stop braking cycles, where high heat inputs over short times produce very high thermal gradients and result in very high residual stresses in the rim surface layers.

The number of braking cycles required to propagate thermal fatigue cracks to a critical length to produce stage 3--overload fracture is estimated to be 10^4 to 10^7 cycles.[3] Many less cycles are required for thermal cracks, although the exact number is not clear.

Thermal cracks have often propagated to a critical length in only a few hundred severe stop brake cycles. However, in extreme cases, thermal crack propagation to complete wheel fracture has been observed to occur after only a single brake application! Again, the reasons for this are not well understood.

Stage III - Wheel Fracture

When the cracks reach a critical length, stress concentration at the crack tip and tangential residual tensile stresses in the rim combine to rapidly fracture the remainder of the rim run by brittle cleavage. Once the crack has penetrated entirely through the rim, its subsequent path depends on the stress state of the plate. In severe cases where the plate is also completely in tension, the crack may simply extend directly through to the hub. More often, however, the crack will turn to fracture circumferentially at the point where the stresses in the plate become compressive. In either case, the usual result is the catastrophic fracture of the wheel into several pieces that fly apart as the stored elastic energy is released. Even if the wheel manages to hold together, its load carrying capability is ruined and a train derailment may still occur.

The critical crack length for the onset of complete wheel fracture is a matter of great concern which, although having received a good deal of study, is still unknown. Based on fracture mechanics and residual stress calculations, it has been estimated to range from 0.2" to 1".[3] However, in wheels where the residual stresses were compressive, cracks over 3" have been found without wheel fracture.[4] Knowing the critical cracks length under any condition would be very useful for maintenance inspection to control wheel life and prevent wheel fracture in service.

The critical crack length depends on two factors: (1) the residual stress level built up in the rim and (2) the resistance of the steel to withstand brittle crack propagation. Which of these factors is of greatest importance in controlling wheel fracture is still debatable. They will each be considered in turn.

RESIDUAL STRESSES

Many feel that the residual stress built up in the rim of the wheel controls the critical crack length for wheel fracture. Assuming a substantial crack already exists in the rim, complete brittle fracture of the remainder of the rim is only possible if a general residual tensile stress condition is present. Since many wheels are manufactured with general residual compressive stresses in the rim, a stress reversal must occur. This is only possible through long drag braking cycles, when a long and steady but lower-rate heat input (such as would arise during braking down a long, gradual slope) produces a significant general heating of the rim. The accompanying thermal expansion causes the rim to yield in compression at elevated temperatures of 200 to 600°C as it is restrained by the colder central plate. Then, during subsequent air cooling, the contraction of the rim around the cold rigid plate produces a stress reversal that results in residual tensile stress, particularly in those surface layers which had yielded in compression. The plate then generally is forced back into compression. The tensile stresses will reach their maximum once temperature throughout the wheel has equalized, which occurs close to room temperature. Thermal crack cleavage propagation and wheel fracture is then likely from those tensile stresses. This mechanism is consistent with experimental observations that crack formation occurs at low temperatures (below 50°C)[1] and is accompanied by an audible "ping" or "clink" sound. Wheels have even been found to explode by themselves several days or even weeks after their last thermal cycle.

Some researchers believe that the residual stress profile is controlled solely by the most severe braking cycle experienced by the wheel.[4,5] This leads to predictions such as the reversal of residual stress from compression to tension occurring if and when a maximum temperature of 300°C is achieved in the wheel during any drag braking cycle (see Fig. 4).[4]

Other experimental work indicates that residual tensile stress builds up with successive drag braking cycles asymptotically towards a maximum level, as seen in Fig. 5.[2] The presence of compressive residual stresses from manufacturing then simply serves to delay the progression to the maximum tensile stress level and thereby prolong service life. This arises when the wheel rim is quenched during manufacturing (to make it shrink and yield in tension) whereupon compressive stresses are subsequently induced in the rim when the plate finally cools and shrinks.

The effect of tempering on the manifestation of wheel fractures depends on the prior heat treatment. In rim quenched wheels (class C), higher temperating temperatures (or longer times) will lower the level of residual tensile stresses in the plate. This will reduce crack propagation into the plate and favor circumferential spreading of the crack once it has penetrated the rim. In fully quenched wheels (class A), higher tempering temperatures will reduce the residual <u>compressive</u> stresses in the plate and make complete fracture through the plate more likely.

There is still qualitative disagreement over the action of residual stresses in wheel fracture. Obtaining accurate quantitative measurements or calculations of these stresses is difficult and will be the subject of future study.

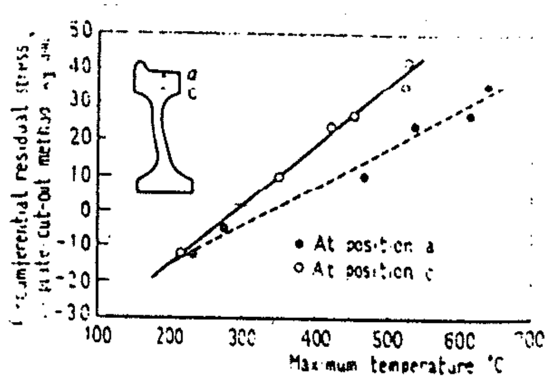


Figure 4 Relationship Between the Maximum Circumferential Residual Stress and Maximum Temperature in the Rim of Wheels[4]

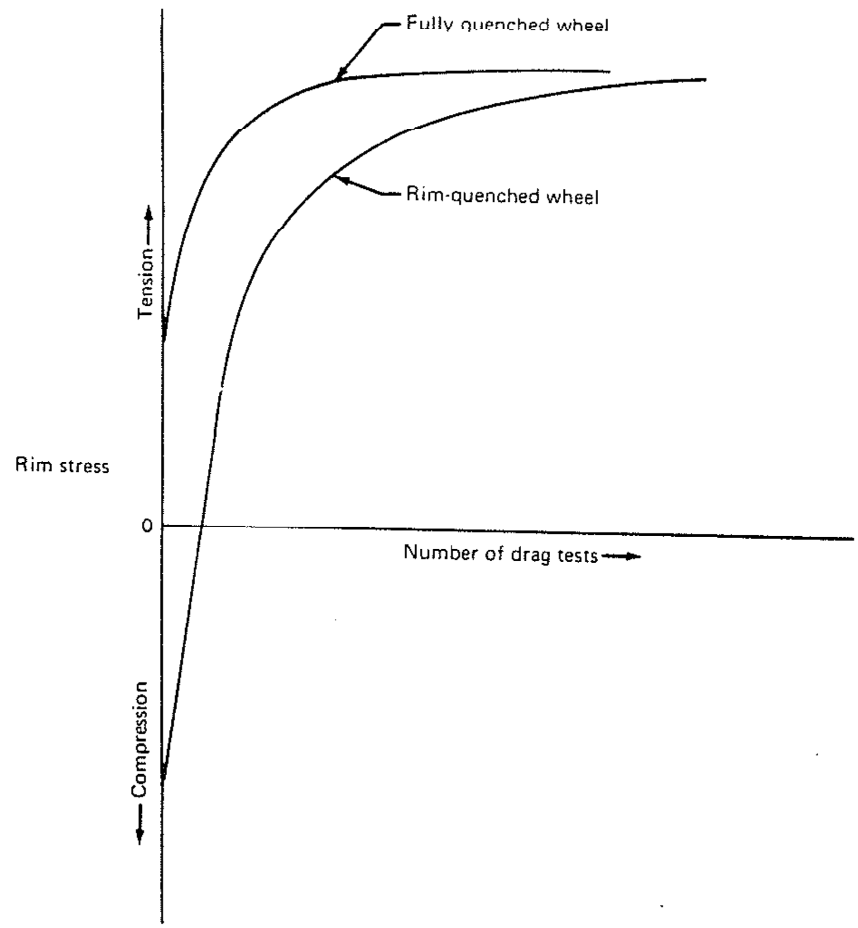


Figure 5 Schematic Representation of Buildup of Residual Stress in Rim by Drag Braking[2]

RELATED CRACKING PROBLEMS IN STEEL

Since relatively few studies are available on wheels themselves, attention will now focus on studies of related cracking problems affecting similar steel grades during other processes. Specifically, this section will review the following three cracking defects affecting semi-finished cast products:

1. Panel cracks in large low-medium carbon, statically-cast steel ingots
2. Cooling clinks in small, alloy-steel ingots
3. Shatter cracks in eutectoid steel blooms for rails

In each of these three processes, internal expansions and contractions due to changing thermal gradients are the sole source of the stresses which, combined with metallurgical embrittlement, lead to brittle fracture. The cracks occur suddenly during cooling, accompanied by very little plastic strain, and are therefore similar to panel cracks.

Panel Cracks

Panel crack formation in static-cast steel ingots is a problem that has plagued the steel industry for several decades. Although the defect is intermittent and affects less than 2% of susceptible steel grades, the problem is persistent and expensive since affected ingots must be scrapped. In a recent study, it was revealed that panel cracks are manifested as two distinct types of cracking problem, referred to as "mid-face" and "off-corner" panel cracks, respectively.[6] Both types of defect affect only killed, aluminum treated steels (0.015 to 0.6% Al) and appear as deep, intergranular cracks that follow prior austenite grain boundaries and are accompanied by a thin ferrite band. They appear to be caused by a combination of metallurgical embrittlement and thermal stresses, primarily caused by the roughly 1% volumetric expansion accompanying the austenite to ferrite transformation.

Off-corner panel cracks run in a discontinuous manner along the edges of the wide faces of large (18,000 to 30,000 kg) steel ingots. They often form rough oval patterns on the ingot wide face and sometimes also on the transverse section. They affect only low carbon steels (0.1 to 0.2% C) with high Mn contents (0.7 to 1.5% Mn) and are particularly likely in micro alloyed steels (containing Nb or V). Sulfur does not appear to be a problem since most of the affected steel grades have low sulfur contents and consequently high Mn/S ratios (exceeding 100). While the exact time of cracking is unknown, they are associated with reheating since off-corner panel cracks are discovered only after removal from the soaking pit. They are also greatly affected by the extent of cooling prior to reheating. Ingots that experience more than two hours of air cooling or are allowed to grow completely cold prior to reheating seldom experience problems.

The results of mathematical simulations suggest that the cracks form in three separate stages.[7] Shallow surface cracks form during air cooling when the surface near the corner falls below the Ar_1 temperature of about 720°C and forms ferrite that shrinks as it cools. Rapid reheating of the ingot surface following a critical amount of air cooling will then result in the enclosure of a region of two-phase steel beneath the surface near the corner that is heating and contracting within a thin zone of retransformed austenite that is expanding. This produces a temporary, subsurface tensile region, whose location near the corner and shape corresponds closely to the eventual location of off-corner panel cracks.

The magnitude of these stresses are calculated to be on the order of only 10 MPa,[7] but this is evidently sufficient to initiate subsurface cracks because the temperature range just above the Ar_1 is susceptible to embrittlement by nitride precipitation at the austenite grain boundaries. Further evidence that this tensile zone is responsible for the second stage of subsurface off-corner panel crack formation is the behavior of this tensile zone with varying track time and reheating practice that corresponds closely to observations of off-corner panel cracks. The tensile zone does not appear with very short track times and with longer track times, it moves deeper below the ingot surface and away from the corner. In addition, the appearance of this tensile zone is delayed and its magnitude is diminished with lower or delayed reheating conditions.

The lack of extensive decarburization found on the crack surface suggests that the cracks do not propagate through to the surface at this time, however. This is because the thin zone of retransformed austenite is both under compression and has better ductility, after trapping the aluminum nitride precipitate harmlessly within new grains. The final stage of propagating the subsurface cracks through to the surface or linking them with existing surface cracks does not occur until the latter stages of reheating or, more likely, during air cooling following removal of the ingot from the soaking pit.

Because off-corner panel cracks affect low carbon steels, are closely associated with reheating and appear to depend on the stress accompanying $\gamma \rightarrow \alpha$ phase transformation, they appear to form via a different mechanism than could contribute to railroad wheel fracture, which occurs at much lower temperatures.

Mid-face panel cracks[6] appear to have somewhat more in common with railroad wheel fracture. These cracks affect only small (1,500 to 6,000 kg) hyper-eutectoid, pearlitic steel ingots and usually exhibit a single, continuous, longitudinal cracks down the center of one of the ingot faces as shown in Fig. 6.[7] They are found exclusively in steels containing 0.4 to 0.75% carbon, although certain alloys containing either 1% Cr, 1% Ni or 1.5% Mn are particularly prone to this defect and are affected at slightly lower carbon contents (0.3 to 0.65%).

Because the incidence of mid-face panel cracking increases with increasing aluminum content (from 0.015 to 0.06%) and nitrogen contents (from 0.004 to 0.012%), embrittlement from AlN nitride precipitation is believed to play a major role in crack formation. Further evidence of this is the beneficial influence of titanium V and possibly Zr in reducing the incidence of mid-face panel cracks. Another interesting observation is that while steels forming both bainite and ferrite (in the form of grain boundary networks) are prone to cracking, fully bainitic steels are not.

The previous study involving mathematical modeling illuminated the mechanism responsible for mid-face panel crack formation.[7] It can be explained using the stress history for two important locations in a 355 mm x 355 mm ingot shown in Fig. 7. During air cooling, the mid-face surface experiences high compressive stresses (reaching 200 MPa) during the transformation from austenite to ferrite and accompanying expansion. Subsequently, this critical location experiences the highest tensile stress of any location in the ingot (approaching 100 MPa) as the surface ferrite seeks to contract while the subsurface (between the Ar_3 and Ar_1) is transforming and expanding.

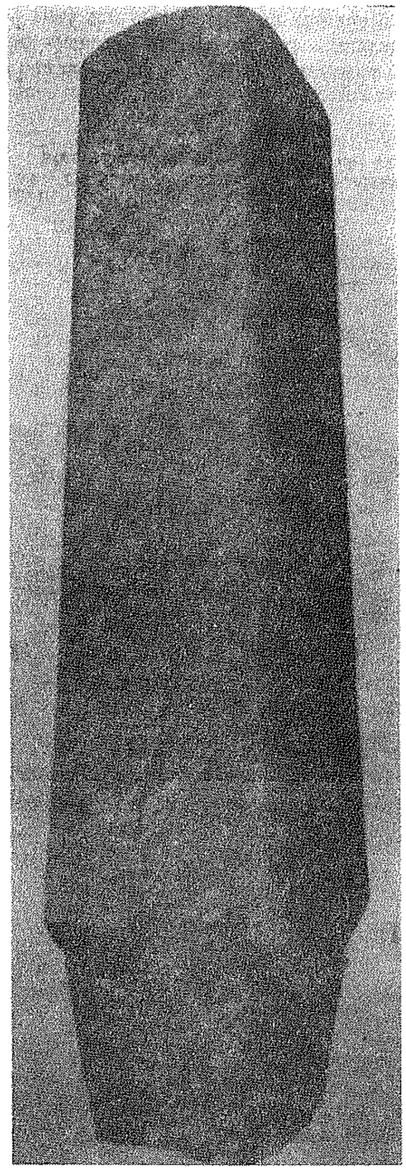


Figure 6 Mid-face Panel Crack in 350 x 350 mm Square En18 (0.4% C, 1.0% Cr) Steel Ingot[8]

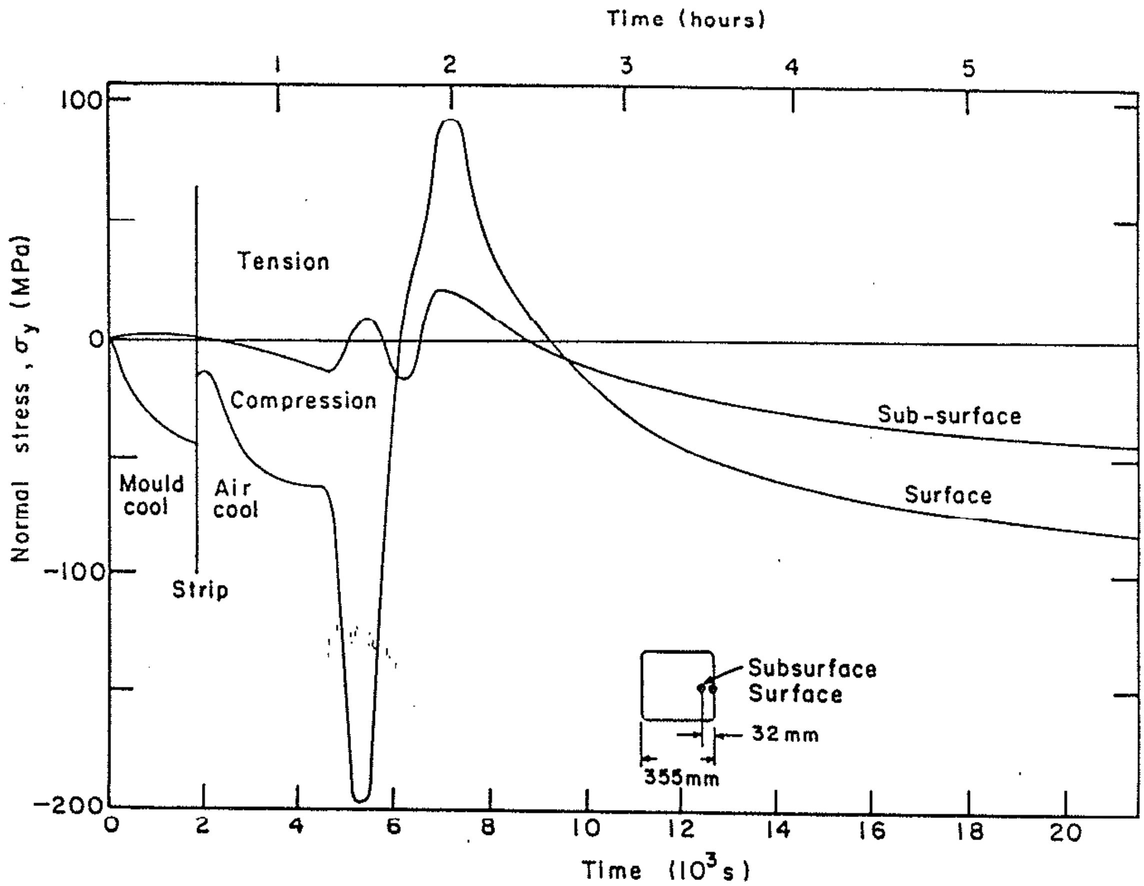


Figure 7 Stress Histories of Two Important Locations in 355 x 355 mm Ingot during Processing

This produces a 1% increase in plastic strain and tensile stresses aligned directly across the grain boundaries, which is the most detrimental orientation for grain boundary fracture. The critical temperature range over which this single, major, tensile peak is experienced is between the Ar_1 and 500°C which falls directly in a lower-temperature zone of embrittlement for steel. In this zone, the pro-eutectoid ferrite networks are surrounded by a hard pearlite matrix which is believed to concentrate strain in the softer primary ferrite located along the prior austenite grain boundaries. In addition, sufficient time has passed for the low strain-rate void coalescence of nitride precipitates in the prior-austenite-grain-boundary ferrite. All of these factors combine to cause mid-face panel cracks when the mid-face surface is between 500°C and the Ar_1.

With further cooling, when the surface goes back into compression (below 500°C), the danger of crack initiation is alleviated but existing cracks are propagated deeper towards the ingot center which ultimately reaches a state of residual tension.

A number of different methods for preventing mid-face panel cracks which are suggested by this mechanism, have met with success in practice. The first of these methods is to prevent the ingot surface from cooling below the Ar_1 and thus avoid the high tensile peak. To ensure that sub-surface cracks also do not form, it would be preferable to prevent the mid-face surface from falling below the Ar_3. This could be achieved by reheating and rolling 355 mm square ingots within the first hour after stripping.

Alternatively, slow, controlled cooling through the critical temperature range might alleviate the problem by delaying and reducing the magnitude of the tensile peak. This agrees with the findings of previous workers that slow cooling alleviated cracking problems.[7] It is interesting to note that the cooling rate experienced by the ingot while the mid-face surface is below 500°C is inconsequential since the surface is in compression by that time. Thus, charging the ingots into a holding furnace for a short time while the ingot surface cools to below 500°C should be sufficient to prevent mid-face panel-crack formation. Longer cooling in the holding furnace is unnecessary. Alternatively, laying one face of the ingot on an insulating surface should also reduce stress generation by concentrating strain in the single hot face above the Ar_3. This would again reduce the magnitude of the tensile peaks experienced by all of the ingot mid-face surfaces and reduce the likelihood of crack formation.

The final solution to mid-face panel cracking is simply to avoid the production of steel compositions susceptible to a low temperature zone of embrittlement. Unfortunately, this requires the lowering of aluminum or nitrogen contents or using an alternative grain refiner. The same mechanism that causes grain boundary embrittlement leading to panel-crack formation is also responsible for the beneficial grain refinement effects so desirable in later processing.[5] Thus, the application of this solution requires the possible acceptance of inferior low temperature properties such as impact toughness.

A comparison of the characteristics of mid-face panel cracks and railroad wheel fracture reveals that these two defects are likely caused by different mechanisms. Railroad wheel steels generally have insufficient aluminum contents to produce significant AlN precipitation and the critical portion of the wheel which propagates the cracks never exceeds the Ar_1 temperature. Nevertheless, the panel crack results indicate that brittle fracture of steel is possible at relatively low strain levels (<2%) and stress levels (<100 MPa) and that the $\gamma \rightarrow \alpha$ phase transformation may produce important alterations in the stress field near the surface of the railroad wheel that have previously received little attention.

COOLING CLINKS[9]

Another serious cracking defect that affects small, statically-cast steel ingots is often referred to as "cooling clinks" because of the sound made when it occurs. The cracks are very straight and transgranular. They usually run transversely across the ingot, often fracturing it into two or three pieces. The cracks are believed to initiate internally and propagate outward during the latter stages of cooling (after the ingot has cooled to about 300°C) or during the first stage of forging when the ingot is placed into the press. They generally affect higher alloy steels such as Hadfield Manganese steel (1.2% C, 12% Mn) or AISI M7 (1% C, 3.5% Cr, 1.5% W, 1.8% V, 8.3% Mo). Steels with high phosphorous contents (above 0.05% P) and high lead contents (above 0.025% Pb) are particularly prone.[9] Another company reports that a variety of steels with carbon equivalents ranging from 0.5 to 1.5% are all susceptible,[10] including En20B which contains 0.4% C, 0.6% Mn, 1% Cr, 0.8% Mo.

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 caused solely by high thermal stresses (no externally applied loads) combined with the inherent low ductility of these steels (possibly produced by their needle-like carbide network structure).

Many of the characteristics of cooling clinks appear similar to railroad wheel fracture:

1. Alloy steels near the eutectoid composition are affected.
2. Cracking occurs during cooling well below the Ar_1 temperature and is accompanied by an audible "ping" sound.

3. The fracture is transgranular and propagates completely through the steel, breaking it into pieces.
4. Cracks are produced solely by thermal stresses and are accompanied by very little macroscopic strain.

Thus, it appears that the same mechanism that produces cooling clinks in statically-cast ingots may also be responsible for fracturing railroad wheels after as little as one severe braking cycle. Unfortunately, the problem of cooling clinks has received very little study. It is simply considered a production cost, or yield loss. Thus, it can shed little light on the railroad wheel fracture problem.

SHATTER CRACKS

Many plain carbon-manganese or alloy steel ingots and blooms are prone to "shatter cracking" or "flaking" during cooling. The problem is particularly common in pearlitic steels used to produce rails. It is dangerous because the cracks are difficult to detect and their propagation can produce sudden failure in service after an unpredictable time delay has passed. The cracks initiate internally (as 1/4 - 1" radial hairline cracks at mid radius)[10] and form at low (ambient) temperatures. They are propagation controlled, exhibiting a transgranular cleavage fracture surface. Alloy steels and steels containing above 0.40% carbon or 1.2% manganese are most susceptible[10] and increasing carbon and manganese contents are detrimental.[11] The effects of various other alloying elements are not well understood, but it appears that Ni, S, P and Al are detrimental while Ce, La, Y are beneficial.[11] The susceptibility of various steels is greatly dependent on the thermal processing route employed. For example, bainitic steels may be more susceptible owing to their low transformation temperatures (~500°C) relative to pearlitic steels (~700°C).[10]

This cracking problem is attributed to hydrogen embrittlement or "hydrogen assisted cracking under a static load."[11] Hydrogen is picked up by the liquid steel during steelmaking from a variety of sources: the furnace atmosphere, rusted scrap, ladle additions, slag moisture, polluted "acid" air, etc. This can result in 5 to 10 cm^3 H_2 dissolved per 100 g Fe.[12] In addition, hydrogen molecules are adsorbed onto the surface of the solid cast steel ingot. They subsequently dissociate into atomic hydrogen, migrate and chemisorb into the steel surface layers:

$1/2\ H_2(g) \rightarrow H$ (dissolved in steel).

The rapid hydrogen diffusion possible at elevated temperatures assists in saturating the entire interior of the steel bloom.

On cooling from austenite to form ferrite/pearlite or bainite hydrogen solubility decreases dramatically (from over 50 ppm at 900°C to 0.35 ppm H at 500°C to only 0.001 ppm at 0°C).[13] At the same time, the rate of diffusion decreases slightly with decreasing temperature, with

D in austenite = 0.00939 exp(-1595/RT) or 0.06 cm^2/s at 1600°F (900°C)

to D in ferrite = 0.083 exp(-11293/RT) or 0.04 cm^2/s at 500°F (260°C).[13,14]

The atomic hydrogen tends to migrate and concentrate in the steel at "traps" such as the voids produced by dislocation pile up and especially in regions under tensile stress. These two factors limit the escape of hydrogen from the ingot surface. Thus, with decreasing temperature during ingot cooling, the large quantities of atomic hydrogen are ejected from solution and combine to form molecular hydrogen, resulting in a pressure increase.

The maximum pressure is reached at an intermediate location between the bloom surface (where diffusion from the surface reduces the H concentration to safe levels) and the center (where higher temperatures allow a greater H solubility).[14] As the internal hydrogen pressure increases, the critical stress for crack initiation decreases. The critical fracture stress is also temperature-sensitive in course pearlitic microstructures (but not in fine pearlite).[15] This increases the likelihood of crack formation to alleviate the pressure.

Once initiated, the cracks propagate rapidly until they are arrested by a region of lower H content or a region of lower tensile stresses. Further crack propagation must await the diffusion of more atomic hydrogen to the site to increase the pressure again by the formation of molecular hydrogen. Hence it can be seen that hydrogen cracking is a discontinuous, time controlled process involving an incubation period prior to fracture plus a finite time for fracture to propagate. Crack propagation also requires a very low strain rate or a static, residual tensile stress state (such as that generated during solidification and cooling of large castings) so that hydrogen can migrate with the crack front. The time for hydrogen cracking to be detectable by ultrasonic testing varies from the almost immediate to several weeks.[10] Thus, a critical combination of tensile stress, hydrogen concentration, and time is required to generate cracks, all of which are greatly influenced by thermal processing conditions.

This mechanism for producing cracks beneath the surface of eutectoid steel rail blooms under the combined influence of hydrogen pressure and thermal tensile stress leads to a number of suggestions for alleviating the problem, which have found success in practice.

The first solution is to reduce the amount of hydrogen initially present in the steel. This can be done most effectively by vacuum degassing the liquid steel while it is in the ladle prior to casting (which allows the majority of the harmful dissolved hydrogen to "evaporate"). Oxygen injection or argon degassing the liquid steel are alternatives. In addition, alloy additions may be roasted, refractories can be kept dry, and every effort made to reduce moisture entrainment at every stage to reduce final hydrogen concentrations. Unfortunately, vacuum degassing and other ladle treatments are very expensive.

A second solution is diffusion treatment of the solid steel whereby dissolved hydrogen is allowed to diffuse from the ingot surface by altering the thermal treatment. For diffusion treatment to be effective, a temperature must be chosen at which hydrogen has low solubility but high diffusivity. Because the diffusion rate in ferrite is faster than in austenite, but the solubility is much lower, the optimum temperature is just below the A_1. The time required for this treatment to be effective increases quite steeply with increase in section size or decrease in temperature.

Thus, one effective practice is to slow cool or hold the ingot/bloom around 650°C prior to reheating, cogging or rolling.[10] An alternative practice is to completely cool the bloom, and then reheat the finished rails to 250 - 550°C to accelerate hydrogen diffusion and reduce the chance of time delayed fracture in service. Longer holding times should be allowed for the controlled cooling of alloy rails or blooms owing to their 25% lower hydrogen diffusivity compared with carbon steels.[13] A third method for reducing hydrogen fracture is to develop compressive stresses in the subsurface layers where hydrogen pressure is highest. Finally, the finished rail should avoid contact with high hydrogen containing aqueous solutions or moist air or polluted, acid containing service environments.

Railroad wheel fracture exhibits several features common to fracture induced by hydrogen embrittlement. In addition to similar alloy compositions being affected, it also exhibits unpredictable, propagation controlled crack growth after a time delay at ambient temperatures. However, railroad wheels are not likely to be as susceptible to this problem as rail blooms for two reasons:

1. Their small size allows for easier diffusion of hydrogen from the wheel surface.

2. Their repeated exposure to heating and cooling cycles involves more available time in the optimum hydrogen diffusivity temperature range and again allows the hydrogen concentration to reduce and thus lowers the likelihood for cracks.

Only at the center of the rim, which has the longest diffusion distance and also remains below the temperature required for rapid hydrogen diffusion during most brake heating cycles, is there any possibility for hydrogen embrittlement to be a factor. This location might correspond to the depth at which a critical crack would propagate to complete wheel fracture, if the hydrogen content was very high and thereby assisted cleavage crack propagation. However, assuming that the initial hydrogen content is low, and the ambient environment is kept relatively moisture and acid free, there should be little likelihood of railroad wheel fracture induced by hydrogen embrittlement.

METALLURGICAL EMBRITTLEMENT

The strains produced by the residual thermal stresses are quite small and generally should produce plastic flow. Fracture can only occur if the steel suffers from some form of metallurgical embrittlement. Atomic bonding theory predicts that very high stresses should be required to produce fracture in an ideal material. In practice, failure invariably occurs at stress levels far below (even several orders of magnitude) theoretical fracture stress, for the following reasons:

1. The crack may propagate via a different mechanism than cleavage such as dislocation movement leading to shear and rupture. This is obviously not the problem for the apparent embrittlement of railroad wheel steel, since these mechanisms produce plastic deformation that would easily alleviate the thermally generated residual stresses.
2. Planes of weakness may exist in the material (particularly along the grain boundaries) where impurity atom segregation and other effects may lower the cohesion strength locally.
3. Finally, the existence of small cracks or notches may act as stress concentrators and increase the local stress level at the crack tip well above that of the average overall stress. This is the phenomenon leading to the development of fracture mechanics and the important material property of fracture toughness (K_{IC}) which will be reviewed for wheel steels in a later section.

The next sections of this report will focus on the metallurgical factors that might contribute to embrittlement over the temperature range and service conditions experienced by railroad wheel steels. The purpose is simply to explore through a literature survey, possible embrittlement mechanisms that have not previously been considered in detail. These include both intergranular and transgranular embrittlement mechanisms operating over different elevated temperature ranges, and reduced fracture toughness (K_{IC}). The importance of these embrittlement mechanisms must be evaluated in future work.

INTERGRANULAR BRITTLE FRACTURE

Steel suffers from several different temperature zones of reduced ductility at elevated temperatures. Figure 8 presents a schematic of the six zones of intergranular fracture and the embrittling mechanism operative in each.[16]

A. Hot Tearing Zone

At temperatures within 40 to 70°C of the solidus, the strain-to-fracture of steel is less than 1%. Many studies have been conducted on this zone of reduced ductility and its embrittlement mechanisms are fairly well understood. As depicted in Zone A of Fig. 8 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 to 70°C of the solidus as shown in Fig. 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.

The transition from brittle to ductile behavior occurs over a narrow temperature range on heating. However, on cooling, ductility may not approach 100% until somewhat lower temperatures are reached. In fact, some embrittlement may be encountered at temperatures as low as 1200°C and, in theory, can continue to operate to as low as the Fe-FeS eutectic temper-

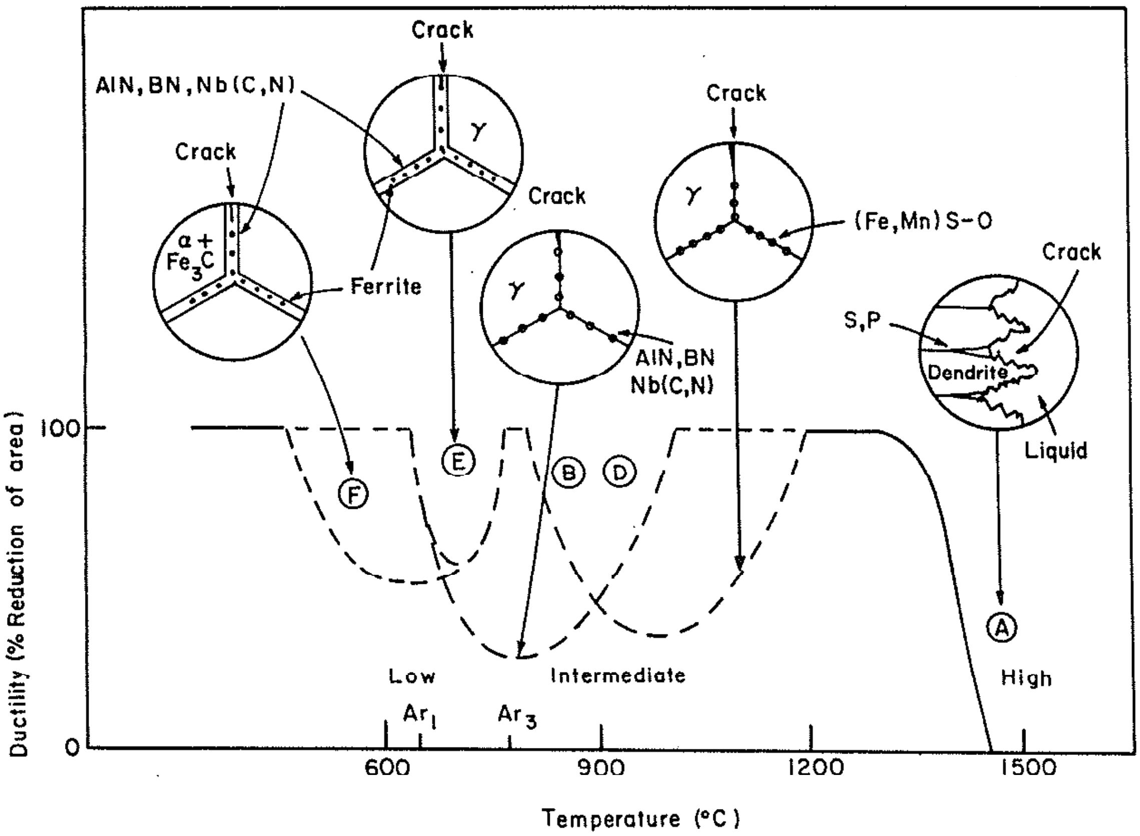


Figure 8 Schematic Representation of Temperature Zones of Reduced Hot Ductility of Steel Related to Embrittling Mechanisms

ature of 980°C. Increased contents of S, P, Sn, Cu and Si all worsen the ductility. Manganese is beneficial since it preferentially combines with S to form less harmful MnS precipitates, thereby preventing liquid film formation. An Mn/S ratio greater than 20 minimizes cracking by this mechanism, partly by raising the ductile/brittle transition temperature. The ductility is relatively insensitive to subsequent thermal treatment and strain rate. However, a comparison with the characteristics of railroad wheel fracture indicates that this mechanism plays no role in the problem.

B-D Intermediate-temperature Embrittlement of Austenite

An intermediate-temperature zone of reduced ductility extends from the Ar_3 temperature to as high as 1200°C. As seen in Fig. 8, it actually consists of several overlapping zones where different embrittlement mechanisms operate. The strain-to-fracture in all cases is considerably higher than for hot tearing, usually exceeding 10% reduction in area. Ductility in this region is controlled mainly by the size, number, and location of precipitates produced by the previous thermal treatment.

At the higher-temperature range of this zone and at high strain rates, (above 10^{-3} s^{-1}), embrittlement is primarily due to sulfide precipitates (principally of iron and manganese). Sulfur segregates strongly and rapidly to the austenite grain boundaries, to form weak sulfide films that induce embrittlement in a similar manner to hot tearing. This can be reduced through high Mn/S levels and by slow cooling or isothermal holds at 900 to 1100°C to allow time for the slow-diffusing Mn to combine with S and encourage less harmful MnS precipitation inside the grains.

At lower strain rates and/or lower temperatures in the zone, nitrides are the principal precipitates responsible for embrittlement. This is because AlN and Nb(C,N) are slow to nucleate in austenite (relative to the easy nucleating, fast-growing, larger sulfides) and produce many, very fine precipitates (<100 nm in diameter). These fine nitride precipitates form preferentially at the austenite grain boundaries. There, they induce embrittlement by enhancing strain concentration and, more importantly, by reducing grain boundary mobility, thereby promoting the nucleation and growth of voids. These creep-type cavities coalesce around the nitride precipitates to produce intergranular fracture. Thus the presence of strong nitride formers such as Al, Nb and B in excess of critical concentrations markedly reduces the ductility. This latter mechanism most likely is responsible for off-corner panel crack formation in large, slow-cooling, low carbon steel ingots. It is not likely to be a factor in railroad wheel fracture, however, owing to the lower temperatures and low levels of nitride forming elements (Al, Nb, etc.) encountered.

E. Intermediate-temperature Embrittlement of Austenite/Ferrite

Another zone of low ductility occurs in the two-phase austenite-ferrite region just below the Ar_3 temperature that corresponds to zone E in Fig. 8. In many respects, it is a continuation of the previously discussed embrittlement zone involving nitride precipitation in single phase austenite. However, the presence of ferrite provides an additional mechanism for concentrating strain at the prior austenite grain boundaries and encouraging intergranular fracture.

At carbon contents approach the eutectoid composition, thin films of pro-eutectoid ferrite form at the prior austenite grain boundaries.

Nitride precipitation is an order of magnitude faster in ferrite, so quickly provides a multitude of sites for the nucleation of microvoids. In addition, the harder and stronger austenite matrix concentrates strain in the soft, grain boundary ferrite networks.

As the temperature decreases, the importance of thermally activated processes involving precipitates decreases. Nitride precipitates coalesce via "Ostwald ripening" and become less detrimental. Thus, the importance of strain concentration in a second grain-boundary phase increases and carbon content becomes more important.

F. Low Temperature Embrittlement of Pearlite

Below the Ar_1 temperature, the ductility of steel is more dependent on the structure present so is therefore greatly influenced by carbon content. In low carbon steels, the phase transformation process actually improves an inherent weakness of austenite by trapping harmful precipitates and voids inside new grains. The new, fully-transformed ferrite and pearlite structure generally has excellent ductility.

However, in steels containing a large fraction of pearlite, a final zone of embrittlement may come into play if the cooling conditions and composition are such that a permanent, thin network of ferrite or cementite forms within a hard, strong pearlite matrix. If this occurs, the strain concentration mechanism described in zone E becomes even more pronounced. Ductility is a minimum just below the Ar_1 when continuous grain boundary networks exist and nitride precipitation is most detrimental. However, the relative embrittlement of susceptible steels can persist to much lower temperatures. This is the mechanism portrayed in zone F in Fig. 8 and is believed responsible for the problem of mid-face panel cracks discussed earlier.

Hyper-eutectoid steels may be particularly susceptible to this embrittlement mechanism when a continuous grain boundary network of brittle cementite is allowed to form. In this case, fracture likelihood increases with pearlite content as the grain boundary networks get thinner and more fragile. The ductility of these steels also depends critically upon the distribution and morphology of the cementite structure. For example, a very high strength, tough and ductile structure can be produced in pearlite through a carefully controlled drawing process. Alternatively, the ductility of the brittle cementite plate structure can be greatly improved by changing its morphology to globules through the spheroidization process.

In conclusion, because this strain concentration mechanism should produce an intergranular fracture along prior austenite grain boundaries that is not observed in fracture of railroad wheels, it is doubtful that this mechanism is very important to the problem. However, it is clear that the inherent brittleness of cementite must play a role in increasing the susceptibility of higher carbon railroad wheels to catastrophic fracture. This will be examined in detail in the later section on fracture toughness.

TRANSGRANULAR BRITTLE FRACTURE

In addition to the many mechanisms causing intergranular fracture, steel is also susceptible to a wide variety of transgranular fracture mechanisms. These arise primarily at lower temperatures (below the Ar_1) and include temper embrittlement, quench aging, strain-age embrittlement, "blue brittleness", dynamic strain age embrittlement and hydrogen embrittlement.

The aging embrittlement processes all raise the yield strength due to the precipitation of carbides and nitrides. This precipitation is caused by time (after quenching), tempering (allowing faster diffusion) or plastic straining. The precipitates tend to move along with dislocations and exert a dragging force on them. The resulting increase in yield strength is accompanied by a decrease in fracture toughness, particularly in the upper shelf. This effect is most evident in the temperature range between 200 and 300 ^{o}C where rapid aging occurs, but can affect lower temperature toughness also.[47] The most important effect is simply raising the ductile / brittle transition temperature. Previous research has focused on these problems in martensitic steels and low carbon alloy steels. No literature on aging embrittlement processes in plain carbon eutectoid steels could be found. Hydrogen embrittlement was discussed in the previous section on clinking cracks.

In order to assess the susceptibility of metals to embrittlement at low temperatures, a number of different tests have been developed to measure the resistance of a metal to brittle fracture. A description of some of the different methods is given in Appendix I.[17] One of these methods measures the property of K_{IC} fracture toughness, which gives a quantitative measure of the critical stress level required to produce fracture. The remainder of this report will review the available literature on the fracture toughness of railroad wheel steels focussing on the important K_{IC} property.

FRACTURE TOUGHNESS

Extensive research has been devoted specifically to understanding the fracture behavior of pearlitic and eutectoid steels.[18-46] However, there is still a great deal of uncertainty regarding the fracture mechanism and controlling factors, and research is continuing. The most important factors affecting fracture toughness are temperature, microstructure, and alloy composition. Accordingly, these variables will be discussed in depth first, followed by other effects such as loading rate, environment, and simple statistics.

To interpret the following discussion in context, it is useful to note the following typical properties of railroad wheel steels and their microstructures:[1]

Ultimate Tensile Strength:	80 - 135 ksi (higher in rim)
Tensile yield strength:	50 - 85 ksi or 350 - 600 MPa
% R. A. :	25 - 50 %
Elongation:	15 - 30 %
Hardness	200 - 300 BHN
Pearlite colony size	13 - 26 μm (increasing from plate to tread)
Prior austenite grain size:	6 - 7 ASTM (finer in tread than rim center)
% ferrite:	10 - 30 %
Pearlite interlamellar spacing:	0.5 - 1.5 μm (coarse)
Cleavage fracture facet size	5 - 150 μm
Ductile Brittle Transition Temp:	0 - 200 ^{o}C (typical about 125 ^{o}C)
Charpy impact toughness	4 - 15 ft-lb (at room temperature)
K_{IC} fracture toughness	30 MPa $m^{.5}$ (at room temperature)
	25 - 35 MPa $m^{.5}$ (at -50 ^{o}C)

Effect of Temperature

The effect of temperature on fracture toughness is great and well-known. Toughness decreases sharply over a small temperature range called the Ductile-Brittle Transition Temperature, or DBTT. Above the DBTT, in the "upper shelf", toughness is generally quite high. It drops roughly an order of magnitude below the DBTT, in the "lower shelf". This trend is illustrated in Figure 9, which also shows the important effect of carbon content. Railroad wheel steels, having carbon contents approaching 0.8%, are generally within the lower shelf at room temperature. Within the lower shelf, the influence of temperature generally is not great. Decreasing temperature in the lower shelf is generally only accompanied by more variability in toughness K_{IC} values.[1] However, one researcher found a slight increase in fracture stress at very low temperatures.[19] The effect of temperature will be examined further in the context of other important variables, specifically microstructure and composition.

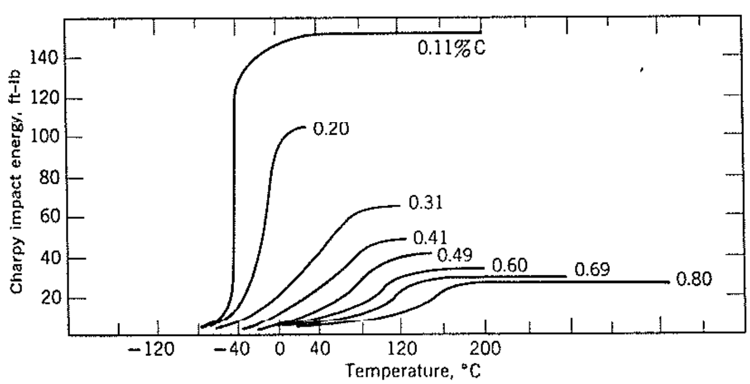


Figure 9 Effect of carbon content on the Charpy impact energy curves for normalized iron carbon alloys.[23]

Microstructural Effects

Microstructure is agreed to be very important in determining the fracture toughness of steel and consequently has received a great deal of study. However, considerable debate still exists over the microstructural parameter(s) which primarily control fracture toughness.[26] Part of the problem is that correlations exist between the microstructural variables, since they are generally impossible to control independently. Strong correlations between mechanical properties also contribute to this confusion, as higher strength steels generally tend to have lower toughness. This is indicated in Figure 10 through the relationship between toughness and tensile ductility. The well-known direct relationship between strength and hardness is illustrated for eutectoid steels in Figure 11,[40] and the inverse relationship between strength and pearlite interlamellar spacing is given in Figure 12.[42]

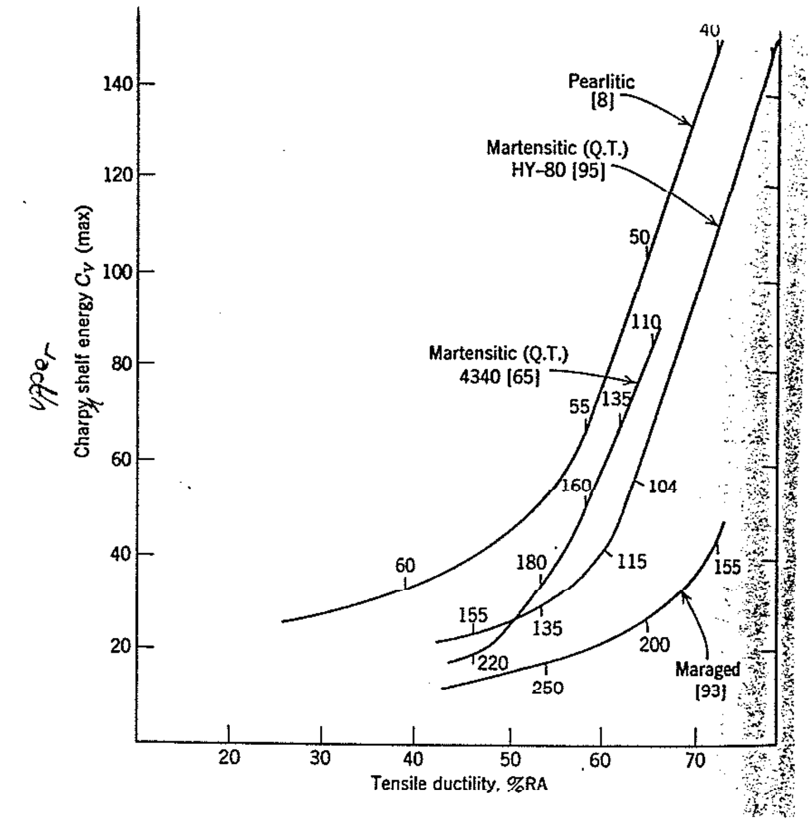


Figure 10 Relation between Charpy shelf energy and tensile ductility for various steels. The numbers adjacent to the curves are the 0.2% offset yield strengths in ksi. These variations in strength, ductility, and tear toughness were obtained by changing the carbon content in the pearlitic steels.[23]

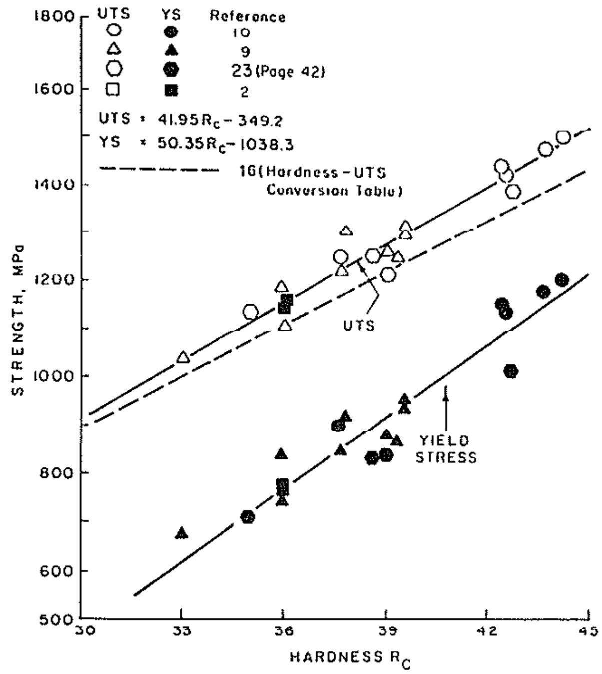


Figure 11 Correlations between Rc hardness and UTS and YS of eutectoid pearlitic steels.[40]

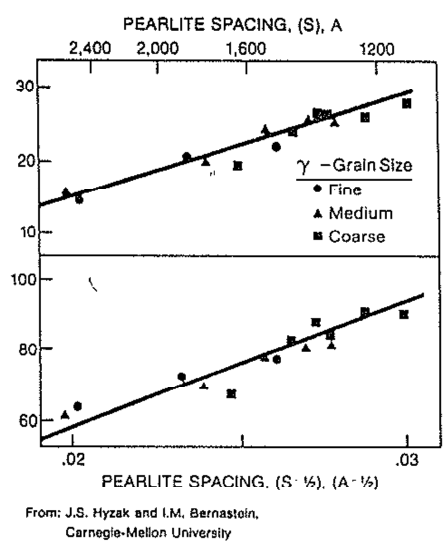


Figure 12 Yield strength and hardness vs pearlite spacing.[42]

The effects of microstructure are most important on the DBTT and upper shelf toughness. They are only of secondary importance at and below room temperature in railroad wheel steels.[30] Several microstructural factors have been suggested to exert a major influence on fracture toughness:

- volume fraction pearlite
- prior austenite grain size
- interlamellar spacing
- cementite plate thickness
- pearlite colony size
- texture

Each of these factors will now be considered in turn.

Volume fraction pearlite: This most important factor is directly related to composition, chiefly carbon content. The morphology of the brittle cementite phase is the critical parameter. As cementite fraction is increased, toughness generally decreases. If the pearlite colonies containing the brittle cementite plates are surrounded by a continuous ferrite network, a dramatic increase in toughness results, in proportion to the ferrite network thickness.[20,23] (This is directly opposite to the effect at higher temperatures near the Ar_1.) Alternatively, if proeutectoid cementite surrounds the pearlite colonies, toughness is even worse. A spheroidized structure is the toughest of all, but has a much lower strength.[23] A bainitic structure is generally intermediate between pearlite and spheroidite in both strength and toughness. At the same strength level, bainite is less tough than pearlite.[23,29] Mechanically, less ferrite results in a smaller tip radius of the advancing crack, which produces more stress concentration and thus lowers toughness.

<u>Prior austenite grain size:</u> Larger austenite grains have been reported by several researchers to reduce fracture toughness in eutectoid steel,[18,20,22] particularly in the lower shelf,[18] and particularly so for K_{ID}, at high strain rates.[18,23] They also increase the DBTT. These effects are illustrated in Figure 13.[28] Increasing prior austenite grain size also decreases Charpy impact toughness for equivalent yield strengths.[28] The suggested reason for this is that prior austenite grain size controls the cleavage fracture facet size. The latter is a strong indicator of toughness since brittle fractures tend to have larger cleavage facets.[18,22] The austenite grain size itself is related to the packet size of the pearlite colonies in the transformed microstructure. The different orientations at the grain boundaries are believed to provide a barrier to crack propagation by forcing more deflections of the crack which results in more facets. This increases the energy needed for fracture in fully pearlitic microstructures.[22,24] However, it may not have a significant effect on the fracture stress.[19] Lewandowski[19] found that large austenite grain size had no negative effect on tensile fracture stress in fully pearlitic steels in the lower shelf below room temperature. The importance of austenite grain size is contrary to hypoeutectoid steels, where ferrite grain size controls fracture toughness consistently.[22]

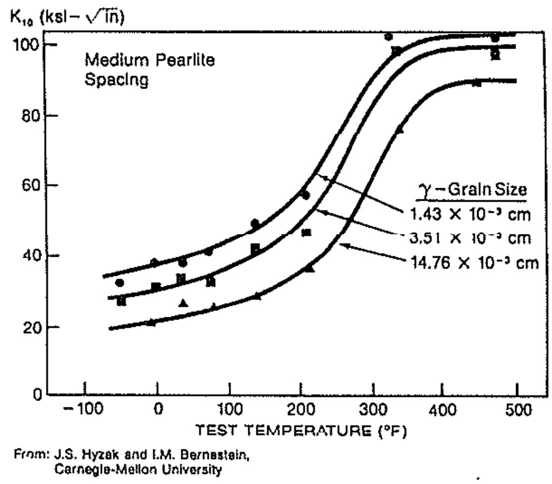


Figure 13 Dynamic fracture toughness as a function of prior austenitic grain size.[28]

<u>Pearlite interlamellar spacing</u>: It is well known that a fine interlamellar spacing correlates with increased yield strength, as shown in Figure 12.[18,40,42] Increased yield strength is generally accompanied by a decreased fracture toughness in all steels.[23] Researchers agree that for a constant strength level, finer pearlite spacing is accompanied by increased fracture toughness and fracture stress. Many researchers also found that the overall effect of finer spacing is an increased fracture toughness in the upper shelf of eutectoid steels.[18,19,23,29] It may also lead to a slight, general increase in dynamic fracture toughness, as seen in Figure 14.[18] However, many researchers report that spacing has no appreciable overall effect on toughness in the lower shelf.[18,29] The mechanism for toughness improvement by finer interlamellar spacing is unclear.

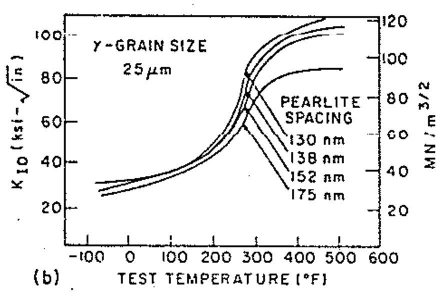


Figure 14 Effect of pearlite interlamellar spacing on dynamic fracture toughness.[18]

Cementite plate thickness: Thinner cementite plates have been suggested to increase toughness.[1,20] However, it is practically impossible to separate this effect from the effect of pearlite spacing discussed above.

Pearlite colony size: Although it has been found to correlate well with cleavage facet size, pearlite colony size has not been found to have an appreciable on fracture toughness. Some researchers suggest that toughness improves slightly with smaller pearlite colonies.[1,20,23]

Texture: The orientation of individual pearlite colonies has been suggested to be an important factor controlling the propagation of brittle cleavage fractures.[20,22,23] Since a change in orientation of the lamellae may interrupt a cleavage burst, any general texturing or preferred orientation of the microstructure may have detrimental consequences on fracture toughness. The cleavage facets may be determined by those pearlite colonies which share a common ferrite orientation.

Since cracks often propagate along [100] planes in ferrite, the ferrite orientations in adjacent colonies may be more important than either pearlite colony size or austenite grain size.[22] This effect is very difficult to quantify since the processing conditions that lead to texturing also change many other aspects of the microstructure.

New process development: The final microstructure developed depends strongly on the thermal processing sequence employed. Austenite grain size is controlled primarily by the austenitizing temperature and pearlite interlamellar spacing can be controlled relatively independently through the cooling rate or isothermal holding temperature after quenching through the Ar_1. New cooling sequences are being developed to refine both the austenite grain size and pearlite spacing and thereby improve both strength and toughness. Austenitizing should be done at low temperatures (about 900 ^{o}C) to achieve the finest grains.[1,29] Then, rapid quenching through the Ar_1 should be followed by isothermal holding during transformation. Finer pearlite spacings are achieved as the holding temperature is lowered. However, to avoid formation of brittle upper bainite, a lower limit exists on the holding temperature. The optimum holding temperature is naturally composition dependent but was found to be about 550 ^{o}C in a high strength rail alloy.[29] A narrow range holding temperature can be achieved practically using a fluidizing bath. An alternative way to control the microstructure is through altering the composition, which will be discussed next.

Effect of Composition

Since fracture toughness is determined directly by structure and only indirectly by composition, it is difficult to isolate individual element effects in the presence of all the other microstructural factors previously discussed. The most important influence of most alloying elements is simply their effect on strength[20], which then is inversely related to their effect on toughness. The reported effects on fracture toughness of each major element in railroad wheel steels will now be examined in turn, keeping in mind that the effect of the element on microstructure is what determines its effect on toughness.

Carbon: Unquestionably, the most important parameter affecting the fracture toughness of steel is the carbon content. Carbon is added to railroad wheel steel primarily to strengthen and increase the hardness and wear resistance. However, by increasing the fraction of brittle cementite present, the fracture toughness is lowered dramatically as shown in Figure 15.[1] As C content increases from 0.4 to 0.8%, K_{IC} decreases by about 30%.[1] This is due to the increased fraction of brittle pearlite. Proeutectoid ferrite softens and toughens the austenite grain boundaries and blunts cracks, particularly when it can form a continuous network. Thus, increasing carbon content above 0.8% further reduces cracking resistance,[36,45] as brittle cementite then replaces ferrite at the grain boundaries. Figure 9 illustrates the effect of carbon on toughness as a function of temperature. In particular, it decreases the upper shelf energy and raises the DBTT.[22] Because K_{IC} is already so low on the lower shelf, it is not affected greatly by increasing carbon content so the increase in DBTT may be of greater importance to railroad wheel toughness.

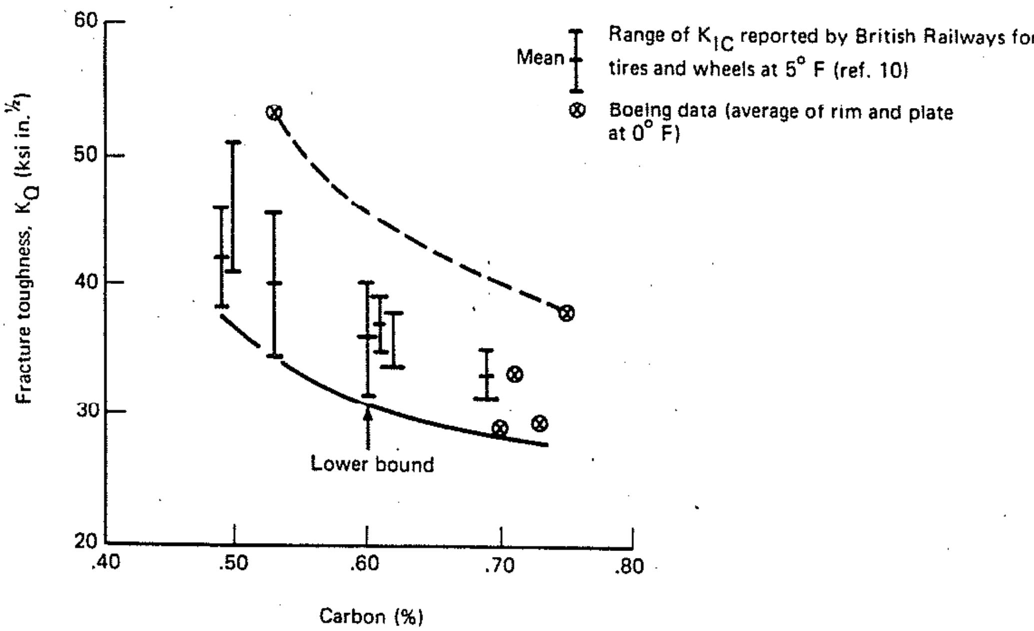


Figure 15 Effect of carbon content on fracture toughness (0 - 5 °F)[1]

Manganese: This element increases steel strength by solid solution strengthening of ferrite[23] and increasing the pearlite fraction.[42] In addition, it refines: the pearlite interlamellar spacing,[20,23,30,42] the prior austenite grain size,[20] and the pearlite colony size.[42] Some researchers find that these effects result in improved toughness in high carbon steels.[30,42] Others find that the net effect on toughness is negligible.[20] Tetelman[23] warns that manganese also increases hardenability so increases the likelihood of forming the brittle upper bainite structure and reducing toughness. However, the improvement in toughness at a given strength level still makes manganese generally a beneficial alloying element for toughness.

<u>Silicon</u>: Like manganese, silicon strengthens steel by solid solution hardening and refines pearlite spacing, producing a questionable effect on toughness. Chelyshev states that up to 0.7% Si improves crack resistance[36], but others find it has no serious effect[33] and may even be detrimental.[20] Bouse[20] found a 1 ^{o}C increase in DBTT per 0.01% increase in Si content but no effect on K_{ID}.

<u>Vanadium</u>: This element is recently being added to rail steels (whose pearlitic structure and composition is similar to railroad wheel steels) as a strengthening agent. It is particularly effective in high N steels, where it produces a 100 MPa increase in yield strength.[38,43] Another study found an increase of 200 MPa in strength and 5 R_C in hardness with 0.1 % V addition.[40] The corresponding increase in hardness is hailed as a method to improve wear resistance and therefore service life. The exact strengthening mechanism is unclear. It includes both solid solution hardening,[20] and dispersion hardening via V(C,N) precipitates which nucleate in austenite.[40] The latter effect saturates for V additions greater than 0.15% as the precipitates coarsen. Fast cooling rates which trap V in solution may be most effective for strengthening.[40] Unfortunately, the increase in strength is accompanied by a decrease in toughness.[20,38] Vanadium additions are believed to lower K_{IC} at -50 ^{o}C, lower tensile ductility, raise DBTT above room temperature, lower Charpy toughness, and possibly introduce undesirable rolling texture.[20,38] It is difficult to determine whether the increase in strength is sufficient to make up for the loss in toughness or whether these effects will be the same in unrolled steel.

Niobium: This microalloy is commonly used as a carbide stabilizer, grain refiner and strengthening agent in low alloy steels. Niobium has not previously been exploited in high carbon rail and wheel steels because of its low solubility and therefore reduced effects (more so than V[40]). Recently, it has received renewed attention because it was hoped to be similar to vanadium in increasing strength and hardness of rails without the detrimental loss in toughness. Cornell[42] suggested that improved room temperature toughness might be achieved with Nb by retarding recrystallization and grain growth during hot rolling and thereby refining the austenite grain size, pearlite colony size and pearlite spacing. However, Glinka found that Nb lowered the toughness of Al-Cr-Nb steels.[31] Its effectiveness as a strengthening agent is also questioned. Hulka found Nb to be very effective at increasing strength and hardness, improving wear resistance of rails by 6X.[34] However, others have found the strengthening effect to be very slight.[40] A final observation is that the strengthening mechanisms of Nb are linked to the high deformation hot rolling process so would not be available in cast railroad wheels.

Nickel: Although Ni has been found to be beneficial in lower C steels, through grain refinement,[23] it has not been found to have an effect in higher carbon steels.[30]

Chromium: Several researchers have reported a negative effect of this hardening agent on fracture toughness[30,31], for example lowering K_{IC} 33 % in Al-Cr-Nb steel.[31]

Aluminum: One researcher found an improvement in toughness in railroad wheel steels after switching to aluminum deoxidation, due to better inclusion control.[41] However, other work suggests that Al may act in a similar manner to Nb, with no obvious bennefits.[31]

Tin: Stephenson found that tin additions below 0.25%, such as might be found as residual impurities, have a negligible effect on room temperature toughness of rail steels.[32] However, Sn may increase the DBTT slightly (1-2 °C per 0.01% Sn) and lower the upper shelf toughness. Like many other alloys, tin increases yield and ultimate strength slightly while reducing ductility.

Nitrogen: Increased N increases the effectiveness of nitride precipitates, principally of V and Nb. It also may increase strength by refining the pearlite interlamellar spacing. However, it generally decreases toughness.

Oxygen: Oxide inclusions act as stress raiser defects if they are large enough, and Fowler[21] found that intermetallic inclusions sometimes initiate cleavage bursts during high growth rate fatigue. However, one researcher reports an improvement in toughness with dissolved oxygen.[39]

Sulfur: Dong[35] attributes premature "pop-in" fracture of rail and wheel steels to the segregation of sulfur. Even at low levels (<0.012% S) homogeneously distributed S greatly reduces fracture of fine-grained plate steel.[39] Little work has been reported on S in pearlitic steels.

Phosphorus: This residual impurity element has been reported to be detrimental to toughness in rail steel.[23,42,45] It acts a ferrite stabilizer, segregates to ferrite grain boundaries and promotes large grains and transgranular fracture. However, its effect is far less than that of carbon. Thus, the importance of P on low temperature fracture is far less than it is at elevated temperatures.

New alloy development: There is a need to improve several different properties of railroad steels simultaneously: fracture toughness, strength (for higher axle loads and speeds) and hardness (for wear resistance). Unfortunately, the latter two properties are generally improved only at the expense of the former, particularly in eutectoid steels. The current philosophy in new alloy development is to improve toughness in the only sure way possible: decreasing the carbon level to reduce pearlite content. The accompanying loss in strength is then replaced through the addition of other alloying elements, that improve strength without affecting toughness significantly. There is still disagreement regarding the best alloy combinations for achieving this goal, however. The most promising elements appear to be those that refine interlamellar spacing or reduce grain size. Likely candidates are manganese and silicon, with V, Nb and Cr being other, more controversial choices. An alternative method for altering the microstructure is through thermal treatment, as previously discussed, or a combination of both.

Effect of loading rate

In a study on ASTM A553 steel, Itoga[27] found that fracture toughness (as determined by critical crack tip opening displacement) is insensitive to loading rate on the upper shelf but decreases with increased loading rate below the DBTT. Similarly, in a study on railroad wheel steels, Carter[1] found K_{ID} (high loading rate) test values lower than K_{IC} values, particularly in the upper shelf.

Effect of residual stress

Theoretically, the presence of residual stress should not effect true fracture toughness. It should simply be accounted for during the calculation of K_{IC} by realizing that the total stress experienced across the specimen is the sum of the applied stress and the residual stress. Neglecting this effect, however, may have serious consequences, since a heat treatment affects residual stresses as well as microstructure. For example, the residual compressive stresses resulting from a quench and temper treatment might erroneously lead one to believe that K_{IC} had improved over that of the untreated, stress free material.

Effect of cyclic loading[31]

Crack propagation during cyclic loading depends on the stress level. At low stress levels, crack propagation is very slow and is controlled by the maximum stress intensity, K_{max}. At intermediate stress, crack propagation is directly proportional to ΔK. At higher stress levels, crack propagation is controlled primarily by K_{max}. The presence of compressive strain between alternate loading cycles may accelerate crack growth.

Glinka[31] reports the interesting observation that K_{max} can slightly exceed K_{IC} during fatigue tests on rail steel. This is thought to be due to cycle strain hardening. The result is unstable crack growth: cleavage bursts followed by periods of crack arrest, similar to that observed in railroad wheels. This same effect causes each crack jump to change the overall crack shape, growing from circular to elliptical as load cycling proceeds.[31] This happens because stress concentration is highest at the deepest penetration of the crack.

Effect of prior strain

Juhas[25] found that prior strain had very little effect on toughness in the lower shelf of eutectoid steels. It caused only a slight increase in both yield strength and K_{IC}, which might explain the increased K_{max} phenomenon observed by Glinka[31] during cyclic loading. Prior strain may decrease toughness slightly in the upper shelf but has no effect on the DBTT.[25]

Effect of atmospheric conditions

Corrosion is also a factor affecting fracture toughness. At high stress concentrations, faster crack growth is produced in moist air than in a vacuum, due to oxidation of the crack front. The important effect of stress corrosion assistance to hydrogen embrittlement has already been mentioned.

Statistical nature of fracture

Brittle fracture is produced by extreme events, not averaged ones, so it is difficult to draw conclusions based on averaged results from different tests. There is always scatter in K_{IC} tests. In fact, variation between laboratories is less than specimen to specimen scatter within a lab.[47] In ferritic and bainitic low alloy steels at low temperatures, it has been determined that cleavage fracture initiates at carbide particles so is necessarily of a statistical nature related to the size distribution of the carbides.[48] Models are being developed based on these probability distributions to extract information on the lower limit of K_{IC} and its overall distribution.[49,50] It is doubtful that this inclusion dependence occurs in railroad wheel steels, but similar phenomena do make the prediction of cleavage fracture very difficult.

Fracture Mechanism

A proposed mechanism for fracture of course pearlitic microstructures is as follows.[17] As deformation proceeds, dislocations are generated at the ferrite / cementite interface which increase the fiber-loading stresses in the lamellae. Fracture of the cementite ensues, followed by intense shear in the adjoining ferrite which fractures adjacent cementite lamellae. Voids then form in the ferrite at the fractured ends of the cementite lamellae, which would continue to grow until ductile fracture. However, local stresses are raised greatly by work hardening. The void growth process is interrupted at some point by another fracture mode: brittle cleavage.

The criteria for the onset of brittle cleavage are not well established. Most work suggests that this occurs when the local stress is raised above a critical fracture stress, such as calculated with the K_{IC} test. Other work indicates that fracture does not always occur in the highest stress region and that a critical fracture strain must also be exceeded.[17] When loading is cyclic, the criteria are even more complex. In either case, the crack can extend by stable cleavage bursts that end at the first unfavorable boundary (such as a grain boundary or change in lamellae orientation).[21]

Most previous work agrees that the cleavage fracture process is propagation controlled. In microstructures containing grain boundary carbides, the critical stage has been proposed to be the propagation of the microcrack into the ferrite matrix.[24] Only for very large grains, (>2 mm), may fracture be initiation controlled.[47] Because crack initiation is not considered important, the presence of inclusions in the steel may not be critical. Lewandowski indicates that inclusions are not important to brittle fracture in course pearlite, so long as they are smaller than the pearlite colony size.[17] Thus, the complete removal of inclusions should not substantially increase the fracture stress.[17]

CONCLUSION

Literature reviews on several different subjects have been conducted that reveal a great deal of information which may be of some relevance to the cracking of railroad wheels. Railroad wheel fracture itself proceeds in three stages: initiation, propagation and fracture. The most critical and least understood phase is the propagation phase, and particularly the point at which a crack of critical length propagates to complete fracture.

A comparison of railroad wheel fracture with other cracking problems affecting steel revealed that cooling clink cracks may form during the cooling of small steel ingots via a similar mechanism that causes railroad fracture. On the other hand, it appears that mid-face panel cracks form during the cooling of steel ingots at higher temperatures (although still below the Ar_1) via a different mechanism. Finally, shatter cracks in eutectoid steel rail blooms form due to hydrogen embrittlement. It appears unlikely that railroad wheels suffer from this problem, owing to their small size and varied thermal treatment. However, this mechanism cannot be ruled out and deserves further study.

Steel is affected by a variety of embrittlement mechanisms that operate in different temperature regimes. Many of these are due to the action of precipitates and result in intergranular fracture along the prior austenite grain boundaries, which is not found in railroad wheels. Several different transgranular fracture mechanisms are also related to precipitate action and/or solute element diffusion. While these could not be investigated in great depth, it is possible that these mechanisms (ie. strain age embrittlement, temper embrittlement, etc.) play a role in railroad wheel fracture. However, they primarily reduce upper shelf fracture toughness to levels that are still well above the lower shelf values of brittle pearlitic steels. Thus, they are likely to be important only if stress levels higher than previously thought are producing fracture significantly above room temperature.

Eutectoid steel has an inherently low fracture toughness. When combined with complex surface loads, cyclic thermal loads, and accumulating residual stresses, it is not surprising that railroad wheels occasionally fracture. To reduce the likelihood of brittle wheel fracture, it is necessary to improve the fracture toughness in the lower shelf. Unfortunately, fracture toughness here is very low and relatively insensitive to most composition and processing variables. Even worse is the increased variation in toughness values as temperature decreases. The statistical nature of cleavage fracture makes its exact prediction very difficult and the prevention of an already rare event even more difficult.

One promising area for a possible solution to the problem is the development of new alloys or thermal processing sequences that would make it possible to produce railroad wheels at lower carbon levels (with a better resulting toughness) while still retaining adequate strength and hardness.

A final suggestion is to return attention to stress generation. The fracture criteria for cleavage fracture may involve both critical stress and strain levels. The thermal cycles experienced during braking produce complex histories of both stress and strain. The accumulation of strain from previous manufacturing sequences and thermal cycles has not previously been considered in great depth. Perhaps it is a critical heat input for a critical time that increases stress and strain at a critical depth in the tread for the crack tip to propagate. The creep strains that occur at higher temperatures are another factor that may be important. The tremendous increase in creep strain rate that occurs in those portions of the tread that exceed the phase transformation temperature has not been previously considered. Residual stresses arise in service from axle and flange loads, in addition to thermal loads. The combined thermal-mechanical loading sequences are another aspect of this area of research.

In conclusion, the two most promising areas for future research to seek understanding and eventually prevent catastrophic fracture of railroad wheels are:

1) develop new alloys and processes that allow lower carbon content and improved fracture toughness

2) renew attention on stress generation (partly through better mathematical simulation)

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APPENDIX I

Several different tests have been devised to measure toughness of metals. The Charpy V-notch test gives a qualitative indication of the relative toughness of different alloys by giving the energy absorbed, the transition temperature and the fracture appearance. However, it yields no indication of the critical stress levels required to produce fracture. The J integral test is valuable for lower strength, higher toughness materials. Only the K_{IC} test (static or low load rate) and K_{ID} test (dynamic or high load rate) give quantitative information on the stress levels at fracture, which can then be used to predict critical stress levels in other geometries. These latter two tests are most valuable for high yield strength, low toughness materials where the speciman remains in a condition of plane strain throughout the test.

In the K_{IC} test, a static tensile load is slowly applied to a pre-cracked specimen until failure. Failure is defined by a 2% crack extension, or a 5% "offset yield" in crack opening displacement. Three empirical rules must be satisfied in order for plane strain conditions to prevail and elastic fracture mechanics to be valid:[17]

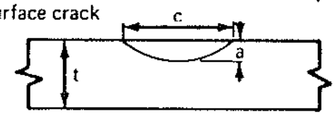
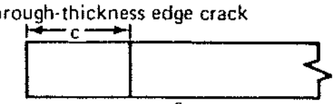
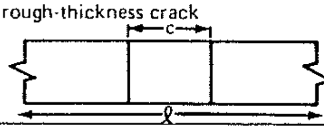
$$1)\quad \text{specimen thickness} \geq 2.5 \left(\frac{K_{IC}}{\sigma_y}\right)^2$$

$$2)\quad \text{crack length} \geq 2.5 \left(\frac{K_{IC}}{\sigma_y}\right)^2$$

$$3)\quad \frac{P_{max}}{P_{(5\,\%\,COD)}} \leq 1.1$$

where: K_{IC} = fracture toughness
σ_y = yield stress
P_{max} = maximum load
$P_{(5\%COD)}$ = load at 5% offset yield

Based on the stress at failure, K_{IC} is calculated from the geometry of the specimen and the crack using equations such as those given in Table 1. These calculations assume that no significant overall plastic yielding has occurred prior to fracture. The "measured" value of K_{IC} can then be used through these same equations to predict the critical stress at failure in components (such as railroad wheels) under similar loading conditions.

Crack configuration	Stress intensity solution criteria for brittle fracture	Remarks
Surface crack 	$K_{Ic} = 1.1\,\sigma\left(\frac{\pi a}{Q}\right)^{1/2}$ (4) $= 0.9\ \sigma\sqrt{a}$ (ref 38)	Solution assumes a/t < 0.5; for deeper cracks, correction factors as required (ref. 39).
Corner crack 	$K_{Ic} = (1.12)^2\,\sigma\left(\frac{\pi a}{Q}\right)^{1/2}$ (5) $= 1.456\sqrt{a}$ (ref. 40)	As above
Through-thickness edge crack 	$K_{Ic} = 1.12\,\sigma\,(\pi c)^{1/2}$ (6) $= 1.98\sigma\sqrt{a}$ (ref. 24)	Solution assumes crack length is small relative to width ℓ; for large c/ℓ, correction factors are required (ref. 24)
Through-thickness crack 	$K_{Ic} = \sigma\,(\pi \frac{c}{2})^{1/2}$ (7) (ref. 24)	As above

σ = Failure stress
Q = Factor that depends on crack shape and ratio of failure stress/yield strength (fig. M8)
a = Critical crack depth
c = Critical crack length

Table 1 Stress intensity and fracture toughness relations for crack configurations experienced in failed wheels[1]