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# Failures Related to Hot Forming Processes

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## Introduction to Failures Related to Hot Forming Processes

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Wrought forms are produced by a wide variety of metalworking operations that can be roughly divided into bulk-working operations and sheet-forming operations. The general distinction here is that bulk working imposes material flow in all directions, while sheet-forming operations are typically limited to two-dimensional deformation. Metalworking operations are also classified as either primary metalworking (where mill forms such as bar, plate, tube, sheet, and wire are worked from ingot or other cast forms) or secondary metalworking (where mill products are further formed into finished products by hot forging, cold forging, drawing, extrusion, straightening, sizing, etc.). These metalworking operations have a twofold purpose. First, they are obviously designed to produce parts with the desired configuration. Secondly, metalworking can develop a final shape with internal soundness and improved mechanical properties by:

- Improved internal quality due to compressive deformation
- Grain refinement and beneficial grain-flow patterns
- Elimination of casting porosity and breakup of macrosegregation patterns
- Improved toughness and/or fatigue resistance due to grain flow and fibering
- Burnished surface and controlled surface quality

However, the beneficial factors of deformation processing can become a potential problem if the process is not well designed and carefully controlled. Potential problems of the deformation process also include:

- *Fracture-related problems:* for example, internal bursts or chevron cracks, cracks on free surfaces, cracks on die-contacted surfaces
- *Metal-flow-related problems:* for example, inhomogeneous grain size; shear bands

and locally weakened structures; cold shuts, folds, and laps; flow-through defects and improper grain flow

- *Poor process control, materials selection, and use problems:* may result in underfill, part distortion, and poor dimensional control; tool overload and breakage; excessive tool wear; poor material use; high scrap loss

The movement of metal during these processes, whether performed at room temperature or at elevated temperatures, makes them common sources of surface discontinuities, such as laps, seams, and cold shuts. Oxides, slivers or chips of the base material, or foreign material also can be embedded into the surface during working. These surface imperfections can produce notches of unknown severity that act as stress raisers, which may adversely affect the performance of the component in service.

Subsurface and core discontinuities may also occur. Subsurface flaws often (but not always) originate from the as-cast ingot due to shrinkage, voids, and porosity that form during solidification. Figure 1 illustrates how rolled bar may contain porosity and pipe imperfections carried over from the cast ingot. These imperfections can serve as sites for crack initiation during fabrication or in service.

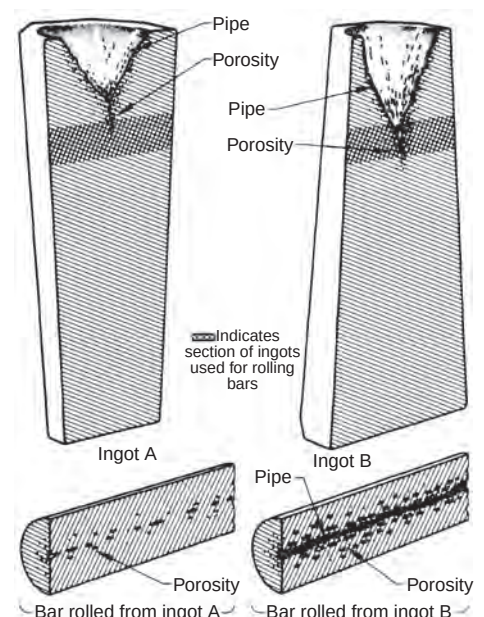
The primary purpose of this article is to describe general root causes of failure that are associated with wrought metals and metalworking. This includes a brief review of the discontinuities or imperfections that may be common sources of failure-inducing defects in bulk working of wrought products.

### Imperfections from the Ingot

Many flaws in wrought products can be traced back to the pouring and solidification of hot metal during production of ingots. For most bulk-forming operations, the starting material is a bar, billet, or flat initially produced from ingots or, in some

cases, made from the product of a continuous casting operation. Primary deformation processes, such as hot rolling, extrusion, and open-die forging, are then used for converting the cast structure to a fine-grained wrought structure while reducing the ingot to a more usable size and shape. Most large open-die forgings are forged directly from ingots, while most closed-die and upset forgings are produced from billets, bar stock, or a preform, all of which have received some previous mechanical working.

The major problems associated with melting and casting practice are the development of centerline segregation and porosity and a condition known as scabs on the surface. Porosity



**Fig. 1** Longitudinal sections of two types of ingots showing typical pipe and porosity. When the ingots are rolled into bars, these flaws become elongated throughout the center of the bars.

is developed in cast ingots when they solidify and is of two types: pipe and blowholes. Factors that work against obtaining a perfect homogeneous cast ingot include:

- Fast shrinkage as the molten metal cools (roughly 5% in volume for steel)
- Gaseous products that are trapped by the solidifying metal as they try to escape from the liquid and semisolid metal
- Small crevices in the mold walls, which cause the metal to tear during the stripping operation
- Spatter during pouring, which produces globs of metal that solidify on the mold walls and ultimately become embedded in the surface of the ingot. These are called scabs, and they usually show up only after rolling or working; as can be expected, they provide a poor surface finish.

Imperfections in the ingot can seriously affect the quality and performance of wrought products. The major ingot imperfections are:

- Chemical segregation
- Ingot pipe, porosity, and centerline shrinkage
- High hydrogen content
- Nonmetallic inclusions
- Unmelted electrodes and shelf
- Cracks, laminations, seams, pits, blisters, and scabs

### Chemical Segregation

The elements in a cast alloy are seldom distributed uniformly. Even unalloyed metals contain random amounts of various types of impurities in the form of tramp elements or dissolved gases; these impurities are also seldom distributed uniformly. Therefore, the composition of the metal or alloy varies from location to location within a cast ingot. Unfortunately, such variation in chemical composition can often be significant and produce deleterious material conditions. This deviation from the mean composition at a particular location in a cast or wrought product is an imperfection termed segregation.

Chemical segregation originates in alloys during the solidification stage. Such deviations from the nominal composition are due to convection currents in the liquid, gravity effects, and redistribution of the solute during the formation of dendrites. Solute rejection at the solid-liquid interface during dendrite formation typically occurs during solidification; thus, a compositional gradient typically exists from the cores of dendrites to the interdendritic regions, with the latter enriched in alloying elements (solute) and low-melting contaminants. Dendrite arms also are generally lower in impurities, such as sulfur and phosphorus in steel, than the interdendritic regions. Consequently, the dendrite arms are stronger and, on working, do not deform and flow as readily as the matrix in which they are incorporated.

Microsegregation characterizes concentrations of elements in interdendritic regions that range in size from a few to several hundred micrometers. By contrast, macrosegregation is the gradient difference, measurable on a macroscale, in alloying elements from the surface to the center of an ingot or casting. Macrosegregation becomes more pronounced with increasing section size. Microsegregation, particularly within secondary arm branches, can be eliminated by homogenization. However, macrosegregation is harder to eliminate because complete homogenization (by high-temperature soaking) would require longer times than are economically acceptable under production conditions. Therefore, in very large sections, gross differences in alloy concentration sometimes persist and are carried into the final product.

One function of primary and secondary deformation is to break up the cast (dendritic) structure and promote chemical homogeneity, and a certain amount of cross-sectional reduction is usually required from the cast ingot to the billet. Hot working can partially correct the results of segregation by recrystallizing or breaking up the grain structure to promote a more homogeneous substructure. Initial working first causes flow in the weaker matrix (interdendritic) regions and tends to reorient the stronger dendrites in the direction of working. With increased mechanical working, the dendrites deform and fracture, thus becoming increasingly elongated.

A certain degree of alloy segregation exists in all wrought products, and hot working can alleviate some of the inhomogeneity. However, if the ingot is badly segregated, hot working just tends to alter the shape of the segregation region into a banded structure. Figure 2 shows banding from a carbon-rich



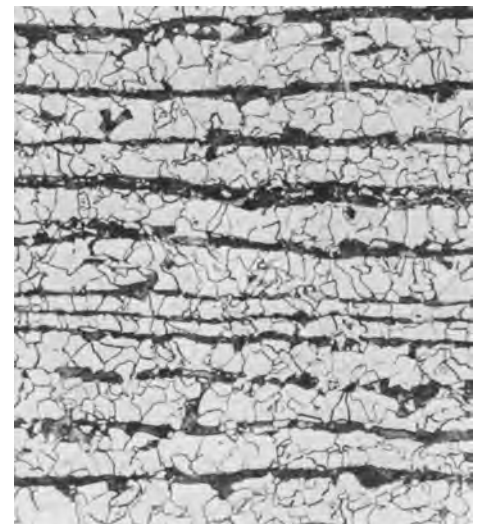
**Fig. 2** Longitudinal section through a hot rolled 1041 steel bar showing a carbon-rich centerline (dark horizontal bands) that resulted from segregation in the ingot. Picral etch. Original magnification: 3 $\times$ . Courtesy of J.R. Kilpatrick

centerline condition in a hot rolled 1041 steel. Figure 3 shows an extreme example of banding in a hot rolled plain carbon steel (1022) in which alternate layers of ferrite and pearlite have formed along the rolling direction. The relationship between increasing percentages of reduction by hot rolling and the intensity of banding in type 430 stainless steel is demonstrated by Fig. 4.

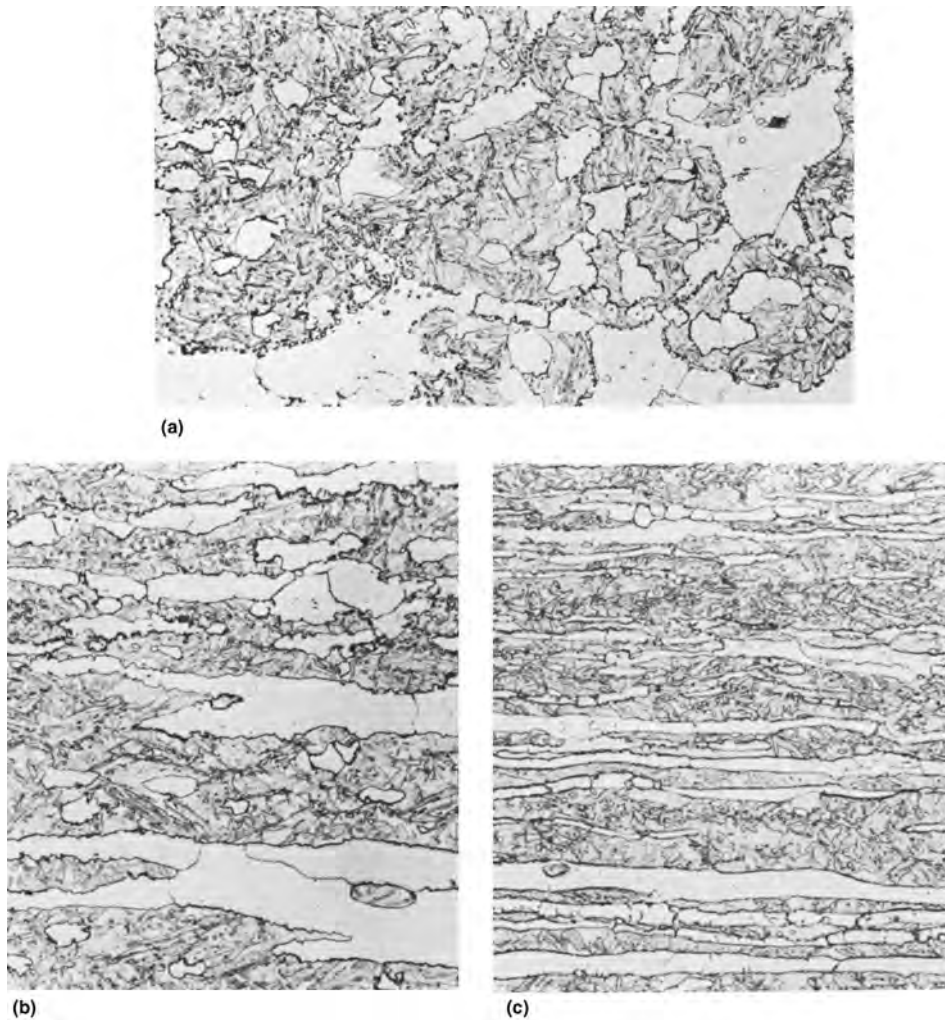
Depending on the type and degree of segregation that develops during solidification, some degree of banding carries over to the wrought form. If banding is severe, it can lead to discontinuities that cause premature failure.

Excessive segregation also can have an adverse effect on subsequent fabrication and heat treatment. In heat treatable alloys, variations in composition can produce unexpected responses to heat treatments, which result in hard or soft spots, quench cracks, or other flaws. Excessive segregation that leads to significant variations in hardness can lead to premature failure and extreme difficulties during cold working or forming. In this case, one of the simplest and most effective tests for incoming material is a simple standard upset test. The details of such a test can be worked out between the supplier and the cold forger.

The methods to reveal the presence of segregation may depend on the alloy and expected impurities. Macroetching is commonly used, and ASTM International has established a graded series (ASTM International E381) of macroetchings for center segregation in steel products. Segregations are revealed by differences in the severity of the etchant attack; segregations at the center may appear as a pipe or may be grouped in some fairly regular form about the center, depending on the shape of



**Fig. 3** Hot rolled 1022 steel showing severe banding. Bands of pearlite (dark) and ferrite were caused by segregation of carbon and other elements during solidification and later decomposition of austenite. Nital etch. Original magnification: 250 $\times$ . Courtesy of J.R. Kilpatrick



**Fig. 4** Type 430 stainless steel hot rolled to various percentages of reduction showing development of a banded structure consisting of alternate layers of ferrite (light) and martensite (dark) as the amount of hot work is increased. (a) 63% reduction. (b) 81% reduction. (c) 94% reduction. Etched with 55 mL 35% HCl, 1 to 2 g potassium metabisulfite, 275 mL H<sub>2</sub>O (Beraha's tint reagent No. 2). Original magnification: 500×

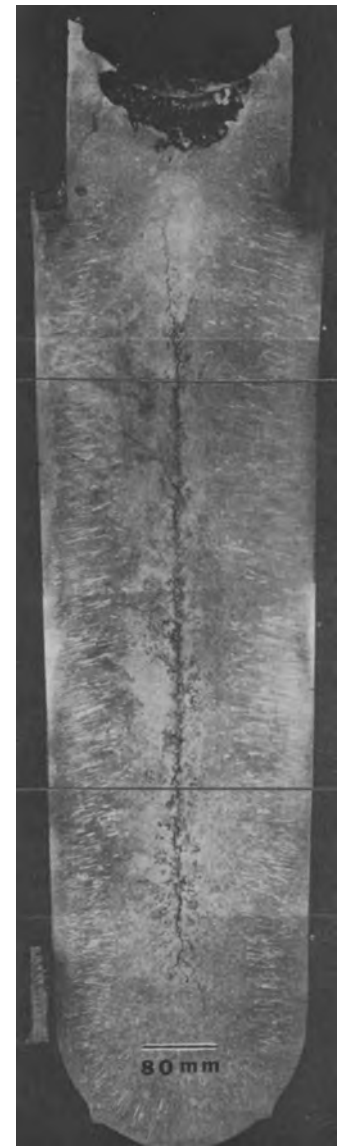
the ingot and the mechanical work done on it. Segregation as revealed by macroetching does not always indicate defective metal. A polished specimen should also be examined under the microscope to determine whether the revealed segregation is metallic or a concentration of non-metallic impurities.

### **Ingot Pipe, Porosity, and Centerline Shrinkage**

A common imperfection in ingots is pipe, which is an internal shrinkage cavity formed during solidification of ingots. It occurs in the upper central portion of the ingot (Fig. 1) during solidification and contraction of the metal, when there may eventually be insufficient liquid metal to feed the last remaining portions as they contract. A concave cavity thus forms at the top of the ingot. The cavity usually forms in the shape of a cone, hence the term *pipe*.

If the ingot is improperly cropped before subsequent working, pipe may be carried through the primary and secondary deformation processes to the finished product. The pipe becomes elongated and is found in the center of the final product, as shown in rolled bar B in Fig. 1. The pipe also can form internal lamination in rolled products. These laminations may not be immediately evident following rolling but may become apparent during a subsequent forming operation. The occurrence of laminations is most prevalent in flat-rolled sheet products.

In addition to the primary pipe near the top of the ingot, secondary regions of piping and centerline shrinkage may extend deeper into an ingot (Fig. 5). Primary piping is generally an economic concern, but if it extends sufficiently deep into the ingot body and goes undetected, it can eventually result in a defective product. Detection of the pipe can be obscured in some cases if bridging has occurred.



**Fig. 5** Longitudinal section through an ingot showing extensive centerline shrinkage

Piping can be reduced by pouring ingots with the big end up, providing risers in the ingot top, and applying sufficient hot-top material (insulating refractories or exothermic materials) immediately after pouring. These techniques extend the time that the metal in the top regions of the ingot remains liquid, thereby minimizing the shrinkage cavity produced in this portion of the ingot. Secondary piping and centerline shrinkage can be very detrimental because they are harder to detect in the mill and may subsequently produce centerline defects in bar and wrought products. Such a material condition may indeed provide the flaw or stress concentrator for a forging burst in some later processing operation or for a future product failure.

Pipe in steel products invariably is associated with segregated impurities, which are

deeply attacked by the acid during macroetching. Cavities in the center that are not associated with deeply etched impurities often are mistaken for pipe, but such cavities usually can be traced to bursts caused by incorrect processing of the steel during forging or rolling. Pipe should be visible after deep etching; it usually can be distinguished from bursts by the degree of sponginess surrounding the defect. Piped metal usually exhibits considerably more sponginess than burst metal. The micrograph in Fig. 6 is an example of secondary pipe, and the micrograph in Fig. 7 is an example of a cavity that appears to be a pipelike flaw.

### High Hydrogen Content

Hydrogen can cause severe embrittlement in steels, especially those with hardnesses above approximately 40 HRC. High-strength, highly stressed parts can crack and fracture as a result of hydrogen embrittlement. Failure by hydrogen embrittlement is even more likely to occur if high residual tensile stresses are present. Hydrogen-embrittlement-related failures are brittle in nature and do not give any prior indication, such as deformation of a part.

The deleterious effects of hydrogen depend on the alloy and the concentration of hydrogen. Hydrogen concentrations in excess of only 3 ppm have been related to the degradation of mechanical properties in high-strength steels, especially ductility, impact behavior, and fracture toughness. Hydrogen concentration in excess of approximately 5 ppm has been associated with hydrogen flakes, which are small cracks produced by hydrogen that has diffused to grain boundaries and other preferred sites (such as inclusions or matrix interfaces). Flaking or hairline cracking in steels, particularly in forgings, heavy-section alloy plate steels, and railroad rails, has been an ongoing problem since the beginning of steel-making and is reviewed in Ref 1.

A major source of hydrogen in certain metals and alloys is the reaction of water vapor with the liquid metal at high temperatures. The water vapor may originate from the charge materials, slag ingredients and alloy additions, refractory linings, ingot molds, or even the atmosphere itself if steps are not taken to prevent such contamination. The resulting hydrogen goes into solution at elevated temperatures, but as the metal solidifies after pouring, the solubility of hydrogen decreases and it becomes entrapped in the metal lattice.

Hydrogen can also be absorbed many places during primary and secondary deformation. It can be picked up during melting, teeming, welding, pickling, or electroplating. If hydrogen is absorbed into the base metal, then it must be removed by a bake-out heat treatment. However, the bake-out heat treatment is only effective for removal of hydrogen absorbed during secondary or postsecondary operations (such as pickling and electroplating) and would not be helpful in removing hydrogen introduced during steelmaking.

Excessive gas contents in ingots are most serious when oxide inclusions are present. Inclusions provide sites for the agglomeration of gas; thus, molten alloys are often treated to reduce the hydrogen content and alkali element concentration as well as to remove nonmetallic inclusions. For example, vacuum degassing is used to remove dissolved oxygen and hydrogen from steel, thus reducing the number and size of indigenous nonmetallic inclusions and the likelihood of internal fissures or flakes caused when hydrogen content is higher than desired.

Despite these methods of producing high-quality steel, problems sometimes arise, particularly if there are no methods available for continuously monitoring metal quality. Hydrogen flaking also is a potential problem for carbon tool steels and for some of the medium-carbon low-alloy steels, such as those used as prehardened plastic-mold alloys. Tool

steel that has cracked will exhibit a coarse, shiny, faceted appearance.

### Nonmetallic Inclusions

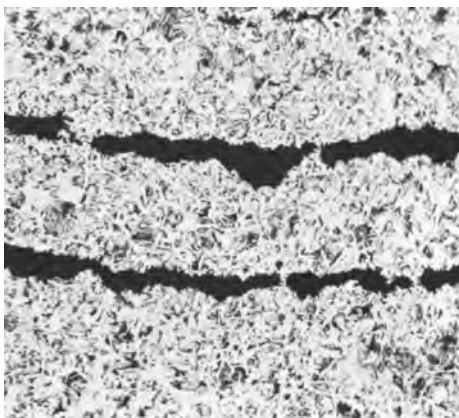
Nonmetallic inclusions are an inevitable consequence of commercial alloys and their respective melting or casting practices. Inclusions that originate in the ingot are carried on to wrought products, even though their shapes may be appreciably altered. Furthermore, additional nonmetallic matter, such as oxides, may develop during intermediate hot working stages and also end up in the finished form.

Two categories of nonmetallic inclusions in metals are:

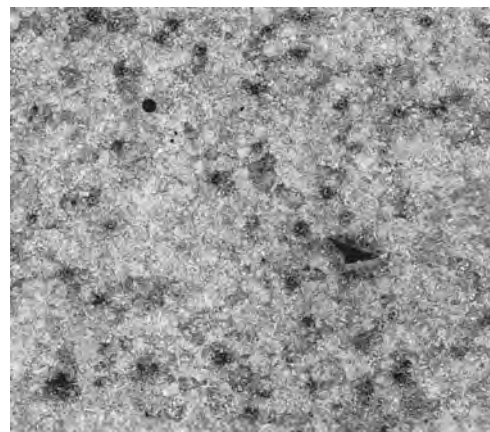
- Those that are entrapped inadvertently and originate almost exclusively from foreign matter, such as refractory linings, that is occluded in the metal while it is molten or being cast
- Those that form in the metal because of a change in temperature or composition

Inclusions of the latter type are produced by separation from the metal when it is in either the liquid or the solid state. Some nonmetallic inclusions form in the liquid before solidification; others form during solidification. In steels, for example, aluminates and silicates generally form before solidification, while sulfides form during solidification. Manganese sulfide inclusions frequently form in the interdendritic regions and primary grain boundaries of steel, where the last of the liquid freezes.

Depending on the alloy, oxides, sulfides, nitrides, or other nonmetallic compounds may occur when solubility limits in the matrix are exceeded. Because these compounds are products of reactions within the metal, they are normal constituents of the metal, and conventional melting practices cannot completely eliminate such inclusions. However, it is desirable to keep the type and amount of inclusions



**Fig. 6** 1038 steel bar, as-forged. Longitudinal section displays secondary pipe (black areas) that was carried along from the original bar stock into the forged piece. Gray areas are pearlite; white areas, ferrite. 2% nital etch. Original magnification: 50×



(a)



(b)

**Fig. 7** Inclusions and a pipelike cavity in tempered martensite of AISI E4340 steel. (a) Original magnification: 100×. (b) Original magnification: 600×. Courtesy of Mohan Chaudhari, Columbus Metallurgical Services

to a minimum so that the metal is relatively free from those inclusions that may cause unwanted discontinuities. Air-melted alloys commonly contain inclusions. Vacuum- or electroslag-remelted alloys more commonly contain nonmetallic inclusions, such as titanium carbonitrides or carbides, when carbon or the hardening element forms precipitate during stabilization and aging cycles. Homogenizing cycles prior to conversion or at an early stage of conversion can be done for re-solution of some precipitates, but not all nonmetallic inclusions are affected by homogenization. In steels, manganese sulfides can be homogenized.

Nonmetallic inclusions are unquestionably one of the most common imperfections involved in failures. Nonmetallic inclusions can easily become stress concentrators because of their discontinuous nature and incompatibility with the surrounding composition. This combination may very well yield flaws of critical size, which, under appropriate loading conditions, result in fracture. Inclusions that are very brittle and tend to fracture and fragment during working operations also can be very detrimental. In fact, with certain refractory and slag inclusions, fragmentation frequently occurs when the amount of cross-sectional reduction in the workpiece is large.

The deleterious nature of nonmetallic inclusions depends on several factors, including chemical composition of the inclusion, volume percentage, shape, orientation, and the mechanical/physical properties of the inclusion as compared to its surrounding matrix (Ref 2). One study has demonstrated that the local constraint of the matrix by hard inclusions during tensile loading can produce severe stress concentrations that depend on the elastic moduli, size, shape, and orientation of the inclusions (Ref 3). For example, inclusions such as manganese sulfide, which are very deformable at hot working temperatures, elongate in the direction of working. This results in an anisotropic material condition which, depending on

its severity, can produce substantial decreases in transverse mechanical properties, such as ductility, fatigue life, and fracture toughness.

The metal-producing industry attempts to control nonmetallic inclusions, but inclusions cannot be removed with complete assurance. In demanding applications, inclusion shape-control methods are used to produce inclusions predominately with a spherical shape, which is less of a stress concentrator with the surrounding matrix and less damaging than inclusions with more angular shapes. The principle of inclusion shape control is to reduce the surface energy between the inclusion and the metal—normally through suitable chemistry modification of the metal. Sulfide inclusions, for example, have been shape controlled by the addition of calcium or rare earth metal treatments. These inclusions are not plastic or deformable at typical hot working temperatures and retain their globular shape. Therefore, they are less injurious with respect to ductility, toughness, and fatigue life in finished wrought products.

### Other Ingot-Related Imperfections

#### Laminations

Laminations in wrought products may occur from various types of discontinuities introduced during the ingot stage. As previously noted, pipe can form into laminations in rolled products. Laminations may also form from spatter (entrapped splashes) during the pouring of the molten metal into the ingot mold. These imperfections are elongated during rolling or other working and are usually subsurface, as shown in Fig. 8(b) for bar stock and illustrates a lamellar structure opened up by a chipping tool.

#### Slivers

Slivers are most often caused by a rough mold surface, overheating prior to rolling, or abrasion during rolling. Very often, slivers

are found with seams. Slivers usually have raised edges, as shown in Fig. 8(c).

#### Scabs

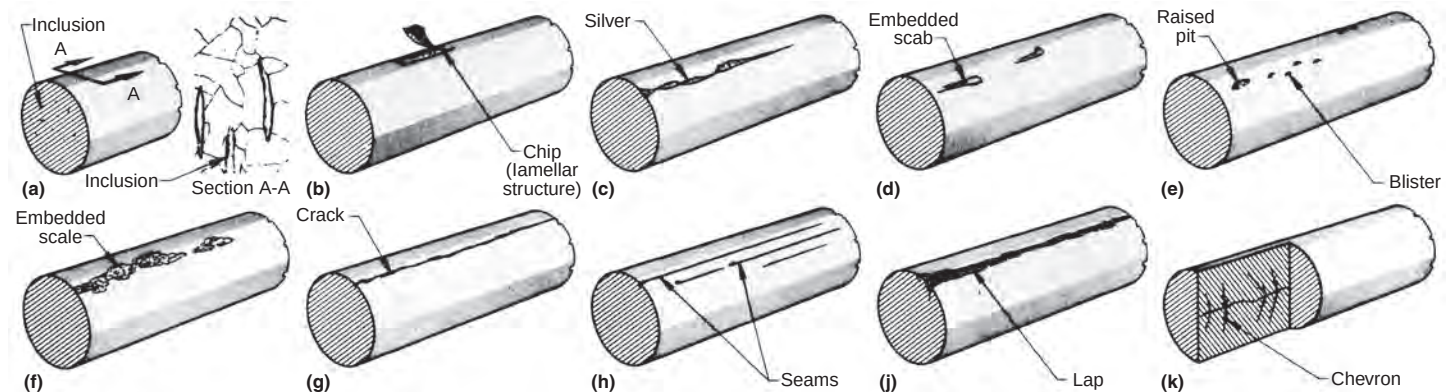
Scabs are caused by improper ingot pouring in which metal is splashed against the side of the mold wall. The splashed material, or scab, tends to stick to the wall and become oxidized. The metal first freezes to the wall of the mold, then becomes attached to the ingot, and finally becomes embedded in the surface of the wrought product (Fig. 8d). Scabs usually show up only after rolling and, as can be expected, provide a poor surface finish. Scabs thus bear some similarity to laminations.

#### Pits and Blisters

Gaseous pockets in the ingot often become pits or blisters on the surface or slightly below the surface of bar products. Pits may also be caused by overpickling to remove scale or rust. Pits and blisters are both illustrated in Fig. 8(e). Embedded scale may result from the rolling or drawing of bars that have become excessively scaled during the prior heating operations. The pattern illustrated in Fig. 8(f) is typical.

#### Cracks and Seams

Cracks and seams are often confused with each other. Cracks with little or no oxide present on their edges may occur when the metal cools in the mold, setting up highly stressed areas. Seams develop from these cracks during rolling as the reheated outer skin of the billet becomes heavily oxidized, transforms into scale, and flakes off the part during further rolling operations. Cracks also result from highly stressed planes in cold-drawn bars or from improper quenching during heat treatment. Cracks created from these latter two causes show no evidence of oxidized surfaces. A typical crack in a bar is shown in Fig. 8(g).



**Fig. 8** Ten different types of flaws that may be found in rolled bars. (a) Inclusions. (b) Laminations from spatter (entrapped splashes) during the pouring. (c) Slivers. (d) Scabs are caused by splashing liquid metal in the mold. (e) Pits and blisters caused by gaseous pockets in the ingot. (f) Embedded scaled from excessive scaling during prior heating operations. (g) Cracks with little or no oxide present on their edges when the metal cools in the mold, setting up highly stressed areas. (h) Seams that develop from elongated trapped-gas pockets or from cracks during working. (i) Laps when excessive material is squeezed out and turned back into the material. (k) Chevron or internal bursts. See text for additional discussions.

### Unmelted Electrodes and Shelf

In consumable electrode remelting, unmelted electrode material and shelf are two other types of ingot flaws that can impair the quality of the resulting product. Unmelted electrodes are caused by chunks of electrodes being eroded away during consumable melting and dropping down into the molten material as a solid. Shelf is a condition resulting from uneven solidification or cooling rates at the ingot surfaces. Unmelted pieces of electrode or shelf conditions appear infrequently in vacuum-melted alloys. Either of these conditions can seriously degrade the product quality. Macroetching or another appropriate type of surface inspection of the machined forging or the billet is the most effective method of detecting unmelted electrodes or shelf conditions.

The conversion practice (primary and secondary deformation) plays an important role in producing a product with a sound center condition. The conversion practice must impart sufficient homogenization to reduce or eliminate the detrimental conditions that are characteristic of the cast ingot. Ingot pipe, unhealed center conditions, voids, and segregation are all melt-related discontinuities that must be mitigated by the conversion practice. Failure to do so may cause the casting flaws to be carried through to the finished product. Normally, it is sufficient to achieve a specified reduction ratio to ensure that a quality product is produced. However, an appropriate conversion practice cannot make up for a poor-quality ingot.

### Primary Reduction Flaws

Flaws can also be introduced during primary and secondary reduction of ingots or billets. Internal bursts will occur, usually on the centerline where the metal is weaker, possibly

from pipe, porosity, segregation, or inclusions. The tensile stresses on the centerline can be sufficiently high to tear the material apart internally, particularly if the hot working temperature is too high. Such internal tears are known as forging bursts, ruptures, or burning in the case of steel forged at too high a temperature. Similarly, if the metal contains low-melting phases resulting from segregation, these phases may cause bursts during hot working of the ingot or billet. Bursts can also occur during subsequent bulk-working operations.

Surface flaws from primary reduction may include:

- *Laps*: These appear as linear defects caused by the folding over of hot metal at the surface. These folds are worked into the surface but are not metallurgically bonded (welded) because of the oxide present between the surfaces. This creates a sharp discontinuity that may result in a failure.
- *Seams*: These are a surface defect that also appears as a linear discontinuity. They occur from a crack, a heavy cluster of nonmetallic inclusions, or a deep lap (a lap that intersects the surface at a large angle). A seam can also result from a defect in the ingot surface, such as a hole, that becomes oxidized and is prevented from healing during working. In this case, the hole simply stretches out during forging or rolling, producing a linear crack-like seam in the workpiece surface.
- *Slivers*: These are loose or torn pieces of steel rolled into the surface.
- *Ferrite fingers*: These are surface cracks that have been welded shut but still contain oxides and decarburization.
- *Fins and overfills*: These are protrusions formed by incorrect reduction during hot working.
- *Underfills*: These are the result of incomplete working of the section during reduction.

- *Rolled-in scale*: This is from an earlier heating operation that was not properly removed.

### Imperfections in Wrought Forms

Various terms are used to describe surface and subsurface imperfections in wrought product. The meaning of terms can vary by location and industry, but Fig. 8 is a schematic illustration of some terms used to describe flaws in rolled bar stock; this schematic is not a complete summary of possible imperfections. Imperfections, whether at or below the surface, can adversely affect performance of a part by creating a notch of unknown severity and serve as a crack-initiation site during fabrication or in service. Corrosion and wear damage can also be assisted by discontinuities, especially at the surface. These flaws may be created from the melting practices and solidification of ingot, the primary or secondary working of the material, or the metallurgical characteristic of a particular alloy system.

It must be clearly recognized that the processed metals described in this article always have some imperfections or discontinuities that can be acceptable if they do not interfere with the utility or service of a part. Laps, seams, bursts, hot tears, and thermal cracks are typically considered to be manufacturing defects. However, whether these manufacturing defects are a contributing factor in an in-service failure is a separate question that would need to be confirmed during a failure analysis. A discontinuity or flaw only becomes a service defect when it interferes with the intended function and expected life of a part. This distinction is important in failure analysis because a discontinuity or imperfection may be present, even though the failure is attributed to a different root cause. Acceptable imperfections must be carefully spelled out in the material purchase specifications.

## Steel Forging Defects

Many flaws in forged products can be traced back to the initial starting material. In the case of large open-die forgings, this starting material is usually cast ingot or conversion products made from a cast ingot. Impression die forgings are usually smaller and are produced from billets or bar stock that were in turn produced from cast ingots. Thus, flaws produced in the casting and conversion processes can be carried through to forgings. In addition, flaws can be introduced while making billets and bar stock. Billets and bar stock are commonly produced from ingots by rolling or open-die forging.

This section addresses only the types of flaws or defects that can be introduced during

the forging process itself. They may be flaws related to metal-flow issues or may be related to poor heating practices or just poor process control. Defects originating in the ingot-casting process are covered in another section of this article.

### Defects Caused by Overheating

In addition to a well-designed forging process, proper control of heating is necessary in order to produce high-quality forgings. A poor heating practice can cause excessive scale, decarburization, hot tears,

and burning. Excessive scale, in addition to causing excessive metal loss, can result in forgings with heavily pitted surfaces. The pitted surfaces are caused by the scale being hammered into the surface and may result in unacceptable forgings. The development of scale during preheating of ingots, slabs, or blooms is almost inevitable due to the required long heating times. Sometimes, descaling operations involving water sprays or preliminary light passes in rolling operations are not totally successful, and scale may get rolled into the metal surface and become elongated into streaks during subsequent processing.

## Burning

A commonly overlooked defect in forging of steel is burning. It is the result of severe overheating and may be found in areas of massive deformation, where the additional heat from deformation increases the temperature to above the melting temperature. When steel is overheated to near its melting temperature, the material that is last to solidify during steel-making will melt first. This material is higher in phosphorus, and the overheating (burning) will leave behind phosphorus-rich globules in the affected area that are readily visible microscopically. If the globules have not been elongated during forging, then they were created during the forging process itself and were not a preexisting condition in the bar stock. Unfortunately, burned forgings are often overlooked because there is no visually observable evidence of burning on the surface of the forging.

While burning may be detected by ultrasonic inspection methods, these measures are normally not taken except for those cases in which the component is considered to be safety-critical. As a result, burning may not reveal itself unless subsequent processing causes a failure or when the forged part fails in service.

Burning only occurs when normal heating practices are ignored and the forging stock is heated to a temperature approaching the melting temperature of steel. This may occur by accident in slot furnaces where there is little in the way of temperature control other than observation by the unit operator. Induction heating systems are usually well designed to achieve and maintain a desired temperature. However, when dealing with incomplete filling of dies, some forge unit operators may choose to increase the temperature of the forging stock rather than the more difficult task of adjusting the closure setting of the forging dies. Raising the forging temperature to facilitate die filling is often the cause of burning.

## Hot Tears

Like burning, hot tears in forgings are almost always the result of overheating to near the melting temperature of the steel. In areas of high deformation, the added heat will cause the steel to rupture under the stress of deformation. In this case, the rupture will manifest itself as surface cracks that are often ragged in appearance and readily visible to the naked eye.

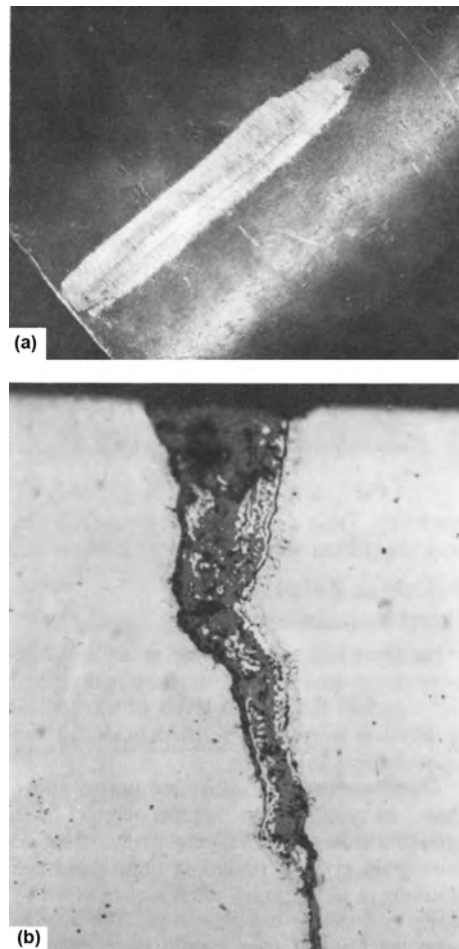
## Thermal Cracks

Thermal cracks may occur as a result of nonuniform temperatures in the forging. Quench cracks are one example of such thermal cracks. Internal cracks, another type of thermal crack, may occur when forgings are heated too rapidly. These occur as a result of unequal temperatures of the surface relative to the center of the mass, and the resulting

differences in the degree of thermal expansion produce tensile stresses near the center. The formation of such cracks depends on both the section size and the thermal conductivity of the material. Large section sizes and poor thermal conductivity promote steep thermal gradients and favor crack formation.

## Imperfections Created during Forging

Many forging defects can be created during the forming process. The defects are almost always caused by improper flow during the forging process. Seams and laps are surface discontinuities that are caused by folding over of metal without fusion. They are usually filled with scale and, on steel components, are often enclosed by a layer of decarburized metal (Fig. 9). Other types of defects are flow-through defects and shear cracks, in which a diagonal shear band is established due to high strain rates.



**Fig. 9** Seam in rolled 4130 steel bar. (a) Closeup of seam. Note the linear characteristics of this flaw. (b) Micrograph showing cross section of the bar. Seam is normal to the surface and filled with oxide. Original magnification: 30×

A special category of defects is the chevron crack. It is a centerline defect that is commonly found in extruded cold forgings. However, it can occur in other cases where high hydrostatic tension occurs at the centerline of a forging.

The defects described in this section are related to the distribution of metal in each step of the forging process. Forgings are often made in multiple steps in order to distribute the metal in a manner to yield a flaw-free forging. Defects can be avoided by proper preform die designs as well as a proper finish die design. Lubrication can also play an important role in the production of defect forgings. Too little lubricant can prevent proper flow and filling of the die. Too much lubricant can collect in die details and cause underfilled forgings.

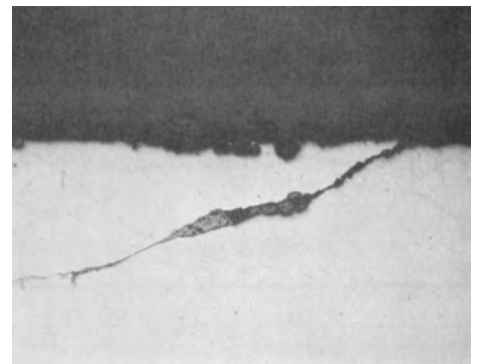
## Lap and Fold Defects

Laps are surface irregularities that appear as linear defects and are caused by the folding over of hot metal at the surface. These folds are forged into the surface but are not metallurgically bonded (fused) because of the oxide present between the surfaces (Fig. 10). Thus, a discontinuity having a sharp notch is created.

A lap or fold results from an improper progression in fill sequence. Normally, a lap or fold is associated with flow around a die corner, as in the case of an upper rib or lower rib, or with a reversal in metal-flow direction. A general rule of thumb is to keep metal moving in the same direction. The die corner radius is a critical tool dimension, and it should be as generous as possible. In progressing through a forging sequence, the die corners should become tighter with each step. Thus, the workpiece fillets are initially large and progressively become smaller as the forging steps are completed. Figure 11 shows schematically a lap forming as metal flows around a die corner.

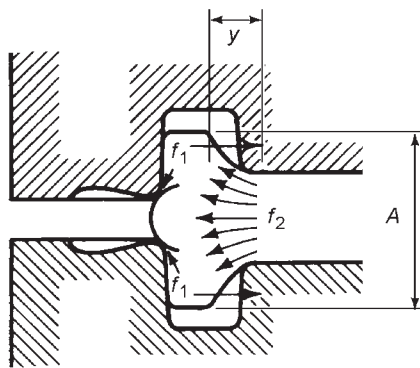
## Flow-Through Defects

Shearing defects are also known as flow-through defects, because they usually result



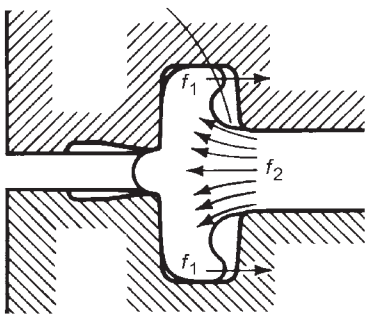
**Fig. 10** Micrograph of a forging lap. Note the included oxide material in the lap. Original magnification: 20×

from excessive metal flow past a filled detail of the part. Flow-through defects form when metal is forced to flow past a recess after the recess has filled or when material in the recess has ceased to deform because of chilling. An example of this is shown in Fig. 12 for a trapped-die forging that has a rib on the top surface. The rib denoted by "2" is filled early in the forging sequence, and significant mass must flow past the rib in order to fill the inner hub, zone "4." The result can be a complete shearing-off of the rib in the worst case, with a lesser case being the formation of a shear-type crack.

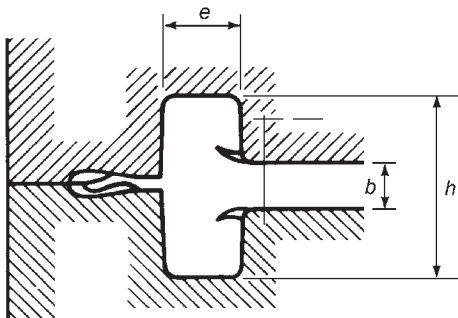


Formation of flash

Curve of natural metal flow



Reverse flow forming a fold



Formed forging defect

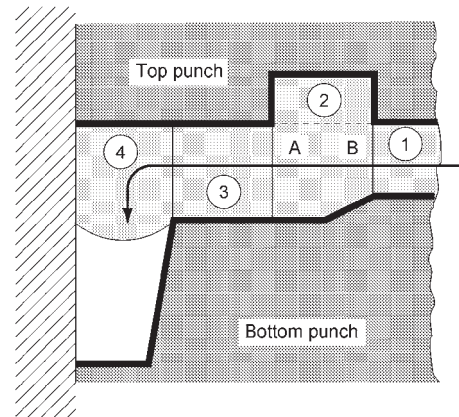
**Fig. 11** Lap formation in a rib-web forging caused by improper radius in the preform die

Similar to laps in appearance, flow-through defects can be shallow, but they are indicative of an undesirable flow pattern and can create a shear band that extends much deeper into the forging. Flow-through defects can also occur when trapped lubricant forces metal to flow past a die impression.

### Seams

Seams are crevices in the surface of the metal that have been closed, but not fused, by working the metal. Shallow seams are often created during the rolling of billet stock that is subsequently cut into mulds and used as forging stock. In fact, it is not uncommon for bar stock specifications to have an allowable seam depth for "forging-quality" steel. Minor seam depths can be acceptable because some amount of the bar surface is lost as scale during the heating process. Minor seams also become unimportant when it is recognized that the forging will be machined on all surfaces prior to use. In these cases, a specified amount of machining stock is added to all surfaces, and minor seams become incidental. However, in the case of precision forging (defined as having some functional surfaces as-forged), seams of any depth are unacceptable, and peeled bars must be specified.

There are several other sources of seams, both mechanical and metallurgical. Seams can develop from cracks caused by working, or seams can develop from an imperfection in the ingot surface, such as a hole, that becomes oxidized and is prevented from fusing during working. In this case, the hole simply stretches out during forging or rolling, eventually resulting in a linear seam in the workpiece surface. Seams also result from trapped-gas pockets, cracks, a heavy cluster of nonmetallic inclusions, or a deep lap. Seams are normally closed tight enough that the seam cannot be visually detected. Seam depths vary widely, and



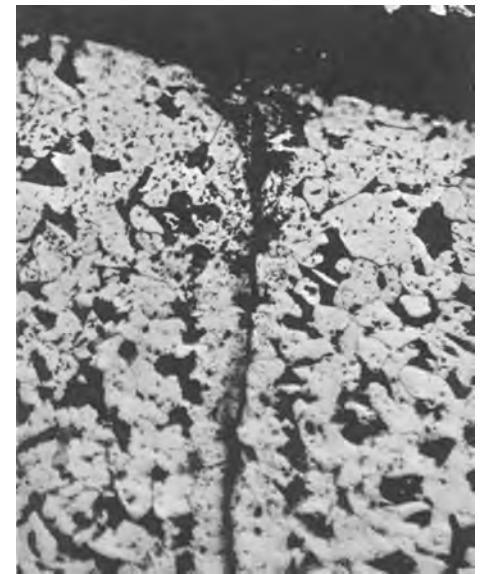
**Fig. 12** Schematic of a flow-through crack at the base of a rib in a trapped-die forging. Excessive metal flow past region 2 causes a shear crack to form at A and propagate toward B.

surface areas sometimes may be welded together in spots. Seams may be continuous or intermittent, as indicated in Fig. 8(h). Oxides in a seam are usually the result of exposure to elevated temperatures during processing. Micrographs of typical seams are shown in Fig. 9 and 13.

Seams seldom penetrate to the core of bar stock. The incomplete removal of seams from forging stock may cause additional cracking in hot forging and quench cracking during heat treatment. They also can be difficult to detect because they may appear as scratches on the forging or because a machining process may obliterate them. Seams may not become evident until the part has been subjected to installation and service stresses, or when the constraint exerted by the bulk of material is removed from the neighborhood of a seam. Sometimes, nondestructive inspection methods such as dye penetrant or magnetic-particle inspection can be used to reveal evidence of seams in both bar stock and forgings.

### Shear Cracks

Shear cracks may also occur in steel forgings. Shear cracks are diagonal cracks usually occurring on the trimmed edges and are caused by very high shear stresses during trimming. They occur when strain rates are high and the heat of deformation is concentrated in a narrow diagonal shear band. Trimming dies used to remove flash must be properly designed and kept in good condition to prevent the formation of shear cracks. Shear cracks can also occur internally in



**Fig. 13** Micrograph of a seam in a cross section of a 19 mm (0.75 in.) diameter medium-carbon steel bar showing oxide and decarburization in the seam. Original magnification: 350×

impression die forgings but are not common in steel because of its relatively high rate of thermal conductivity.

### Centerline Bursts (Chevron Cracking)

Chevrons are internal flaws named for their shape (Fig. 8k, 14). A central burst, or chevron crack, is most commonly associated with extrusion and drawing operations, although it can be generated by forging and rolling processes as well. Central bursts are internal fractures caused by high hydrostatic tension. The severe stresses that build up internally cause transverse subsurface cracks. Some of the factors that can contribute to the formation of chevrons are:

- Incorrect die angles
- Either too great or too small a reduction of cross-sectional area
- Incomplete annealing of slug material
- Excessive work hardenability of the slug material
- The presence of an excessive amount of centerline flaws and other imperfections in the slug materials
- Segregation in a steel slug that results in hard martensitic particles in the center of the slug, which act as barriers to material flow
- Insufficient die lubrication

The conversion practice to bar or billet must impart sufficient homogenization or healing to produce a product with sound center conditions. An example of an unsound condition that did not heal is shown in Fig. 15. A change in deformation-zone geometry is usually sufficient to eliminate the problem. The conservative design approach is to ensure that no hydrostatic tension develops. Often, however, the part or tooling design cannot be changed sufficiently to eliminate hydrostatic tension; if the level of hydrostatic tension can be kept below a critical level, then bursting can likely be avoided. This may be accomplished by a change in lubricant, die profile, temperature, deformation level, or process rate.

Macroetching and ultrasonic inspection methods are the most widely used for identifying regions of internal unsoundness. Bursts usually display a distinct pattern of cracks, are readily detected by macroetching, and do not show spongy areas—thus, distinguishing them from pipes. Figure 16 shows a large burst

that occurred during the forging of an electroslag-remelted ingot. The cause was traced to a weak solidification plane near the bottom of the ingot combined with higher forging temperatures than normal.

### Dimensional Imperfections Unique to Forging

Because of the nature of impression die forging, poorly maintained forging cells may result in defective forgings due to unacceptable dimensional errors. A successful forging operation relies on bringing the two die halves together precisely and with sufficient energy to accomplish the desired deformation. Sometimes the velocity of the moving die is considerable in those cases where velocity is the major energy component (i.e., power hammer

or drop hammer forging). Thus, the act of bringing the die halves together precisely can be a daunting task. As a result, the dimensional specifications for forgings will often take into account the difficulties in achieving some dimensions.

### Mismatch

Mismatch occurs when the die alignment is offset horizontally, as illustrated in Fig. 17. While some mismatch is expected due to the nature of the forging process, the mismatch may become exaggerated when the alignment guides on the forging machine are poorly maintained. Mismatch may also occur if the operators have done a poor job of aligning the dies during the initial setup.

### Closure

A closure dimensional error is illustrated in Fig. 18. For many forging operations, it is usually not possible to bring the die halves into contact with each other to complete the forging. As a result, it is necessary to design the dies and the process so that the desired thickness dimensions are achieved at a slight separation of the dies (usually the amount of the flash gap). Because forging is such a complex process, achieving this precise die position is difficult. Most forging specifications make an additional dimensional allowance for the thickness of the forging to take this into account. Even so, recognizing that undersized forgings are worthless, forgers will often err toward making the forgings oversized,



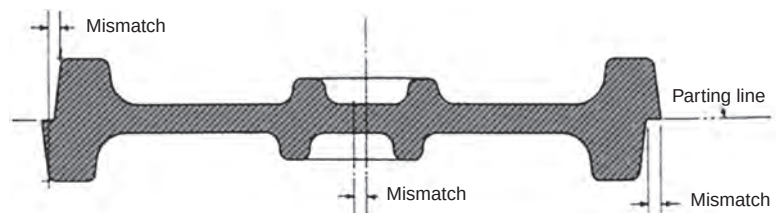
**Fig. 15** Section through a heat-resistant alloy forging showing a central discontinuity that resulted from insufficient homogenization during conversion. Step machining was used to reveal the location of the rupture; original diameter is at right.



**Fig. 16** Cross section of a forged bar showing a forging burst. The burst is located approximately at the centerline of the workpiece. Arrow indicates the direction of working.



**Fig. 14** Centerline chevron cracks in an extruded bar. Courtesy of Bharat Forge Limited, Pune, India



**Fig. 17** Illustration of mismatch in a forging due to poor die alignment



**Fig. 18** Illustration of a closure error where the dies have not been brought together to the appropriate thickness dimension

sometimes even exceeding the extra closure tolerance allowed.

### Underfill

Underfill of a forging is a condition of incomplete capture of the details in the die as specified in the forging drawing. Fundamental to developing a quality forging is the necessity of both the forger and the buyer to agree on a forging design that is economically producible. An inexperienced forging buyer would like to have a highly detailed forging with sharp corners and tight radii. A highly designed forging such as this is usually not necessary and may present serious problems for the

forger. A forger who agrees to produce such a forging is inviting disaster. However, even a properly designed forging can be produced with underfill if the forger does not have a well-designed and well-controlled process. Some of the things that can cause underfill in these circumstances are insufficient material volume, low material temperature, poorly heated dies, and an inadequate lubrication system.

## Nonferrous Forging Defects

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Tim McHenry, Collins Aerospace

Defects found in nonferrous forgings stem from the same general sources as in ferrous forgings, with notable differences in chemical reactivity, melting temperatures, and shear flow characteristics. In all applications, control of casting and pouring practice of ingots is essential to controllable final products.

### Titanium

In the production of a titanium ingot for commercial and aircraft applications, care must be taken to avoid discontinuities related to the solidification process, inclusions, segregation, porosity, and the fabrication processes (rolling and finishing), specifically, avoiding type 1 and type 2 defects (Ref 4).

Depending on the size of the ingots, segregation of elements can lead to structures that cannot successfully be eliminated through heat treatment. The following two types of aircraft-grade materials are not acceptable:

- *Type 1 defects:* regions of interstitially stabilized  $\alpha$
- *Type 2 defects:* high-aluminum defects caused by segregation of  $\alpha$  stabilizers

These conditions can be detected by macro-etching or blue etch anodize techniques. To minimize these conditions, aircraft-grade materials are often triple-vacuum melted to guarantee the quality of the ingot. Stringent nondestructive testing procedures must be employed to ensure these defects are not present.

For titanium forgings, precise temperature control and controlled strain rates are necessary to produce a fully recrystallized  $\alpha + \beta$  structure with uniform grain structure. Nonuniform strain rates can produce an unrecrystallized duplex structure, which, if undetected, can impact the part properties. Sonic inspection and metallographic inspection of the billet is used to determine this condition.

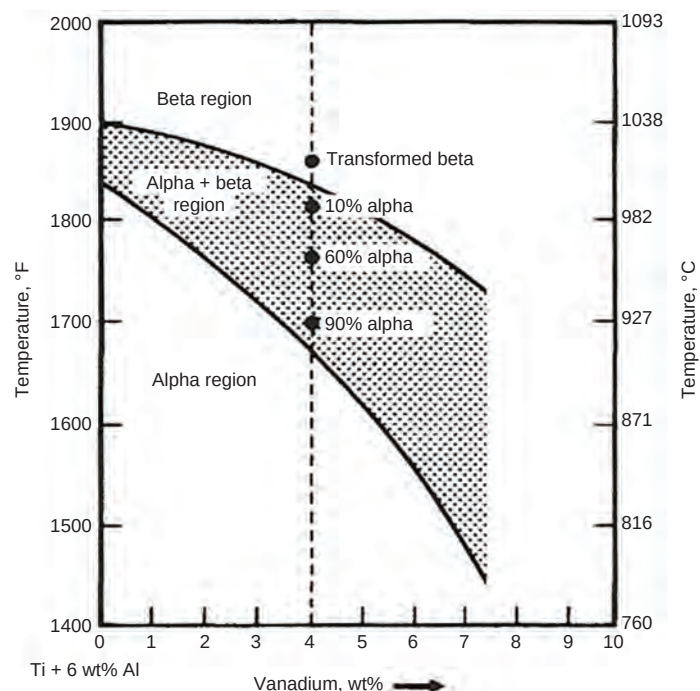
Titanium, like aluminum, has a strong affinity for oxygen. Billets in air at elevated temperatures will develop an oxygen-stabilized  $\alpha$  case, which is brittle. If this case is not removed, it can become entrained in the final product, resulting in crack-initiating defects and reduced fatigue resistance in service.

### Example of Fatigue Cracking

On July 19, 1989, United Airlines Flight 232 crashed while attempting an emergency landing at the Sioux City, Iowa, airport. The National Transportation Safety Board report, AAR-90-06, attributed the crash to a fatigue

crack of a stage 1 fan disc. The crack was initiated at an oxygen-stabilized  $\alpha$  inclusion attributed to the initial billet melting practice.

In production, billets are saw-cut to the proper size to fill the volume of the die cavity and prepared by removing all burrs and breaking edges to ensure smooth metal flow in the die. Preheating of the billet is done in an atmosphere furnace, or it is coated with a glass substance to minimize the infiltration of oxygen and the formation of a stabilized  $\alpha$  layer during heating to forging temperature. The forging process combines mechanical working and a specific thermal cycle to create a product shape and microstructure for the service conditions. The conditions are controlled to create the



**Fig. 19** Phase diagram that predicts the results of heat treatment or forging practice. Source: Ref 5

specific microstructure necessary for the application, be it strength, fracture toughness, or notch sensitivity.

The key forging processing parameter is the  $\alpha$ -to- $\beta$  transus temperature. Two approaches to forging titanium and titanium alloys are to hold the forging temperature 10 to 25 °C (18 to 45 °F) below the  $\alpha$ - $\beta$  transus ( $\alpha$ - $\beta$  forging) and above the transus ( $\beta$  forging) (Fig. 19). The transus temperature is determined by chemistry and supplied by the mill.

Titanium and titanium alloys are highly strain-rate sensitive. The selection of forging method and control of die temperature are critical to the creation of a shape with the appropriate microstructure and mechanical properties.

Press forging methods using controlled die temperatures (isothermal forging) are often used in preference to hammer forging. The more uniform strain rate, along with proper die design and management of lubrication, can optimize material flow conditions by taking advantage of the flow stress/strain-rate sensitivity of titanium. The use of more conventional forging techniques with colder dies and faster deformation rates, in combination with the poor coefficient of thermal conductivity in titanium, can lead to nonuniform deformation and the formation of laps, underfill, and cracks.

Rapid forging processes (such as hammer forging) can also cause adiabatic heating, which, if not controlled, can result in rapid grain growth, change in microstructure, and (in worst cases) insipient melting. The compromise drawn between press and hammer forging is that in press forging a uniform microstructure can be developed that may not be fully recrystallized. In addition, because of the increase in flow stress at higher rates of deformation, greater forging capabilities are needed to fully develop part geometries and to ensure proper grain refinement and final properties for service applications.

The implications of process-related defect formation are discussed in detail in the selected references listed at the end of this article.

## Aluminum

Because of their low density and high potential strength, aluminum forgings are found in automotive and aerospace applications. Like titanium, aluminum is highly strain-rate sensitive. In production with high-strain-rate tools such as hammers and mechanical presses, the rapid delivery of energy can cause as much as a 70% increase in flow stress of such alloys as 6061 (Ref 6). The decrease in mobility of the metal in a die can lead to deformation heating, laps, seams, and cracks in the product. In high-production shops, rapid forming rates can lead to high residual stresses in the product, requiring immediate stress-relief treatments after forging. Further discussion of these topics can be found in the article "Forging of Aluminum Alloys" in *Metalworking: Bulk Forming*, Volume 14A of the *ASM Handbook*, 2005.

In production situations where slower rates of deformation are possible (press forming), dies are heated to near-forging temperature to minimize heat loss during forming. The low melting temperature of aluminum allows for near-isothermal conditions to be established. In some cases, the tool temperature is within 120 to 150 °C (215 to 270 °F) of the workpiece (Ref 7).

## Common Forging Defects

Blisters, which are usually caused by a reaction with the furnace atmosphere, have a white appearance under the blister. Trapping and reacting with oil or lubricant can be caused by uneven metal flow, resulting in a black or dark-gray color under the blister artifact.

Deformation heating (insipient melting) can occur during rapid strain-rate forging. Localized heating impacts the final microstructure and serviceability of the part. The effect is most often seen at triple points in the microstructure.

Overaging of alloys in the T5 condition is another common forging defect. Improper cooling from the forged condition or prolonged exposure at elevated temperatures during heat treatment cycles can result in the agglomeration of precipitates in the structure and a loss of strength. This condition can be determined by conductivity measurements or tensile tests.

Forging lubricants left on parts can create a corrosive environment, which causes in-process pitting corrosion. Lubricants should be removed from the parts after forging, especially in high-humidity environments.

Large grain size can be caused by high initial forging temperatures—typically within 35 to 93 °C (65 to 165 °F) of melting temperature—or insufficient deformation during forging. If the work is subjected to nonuniform deformation or is worked at a rate and temperature outside the optimal conditions for dynamic recrystallization, a texture is formed (Ref 6, 8). Nonuniform grain size will reduce the ductility and strength of the part.

### Example of Forging from Mechanical and Hydraulic Presses

In an experiment, an aluminum alloy aircraft part was forged from the same heat of material using a mechanical press and a hydraulic press. The deformation-rate ratio of mechanical to hydraulic is approximately 10:1. The parts produced on the hydraulic press exhibited large, nonuniform grains insufficient to meet grain-size requirements. The mechanically produced part exhibited a uniform, fine-grained size.

## Copper and Copper Alloys

Copper and copper alloy forgings have the same defects as all forgings—laps, seams, dead zones, and entrained surface contamination.

As with all forging, billet care should be taken to ensure clean starting conditions. Billets are sheared or saw-cut to the proper volume for the application. Burrs and rough edges are removed to minimize the potential nucleation of defects during the forging process.

Heating time for copper alloy billets is critical. Prolonged heating time in excess of what is necessary to obtain uniform billet temperature can cause excessive grain growth and scale. Brass alloys (C37700) are the least sensitive to heating time, but for alloys such as aluminum bronze (C60800), naval brass (46400), and high-purity coppers (101000), time is more critical. In forging of high-purity copper, minimum heating times and rapid cooling after forging are necessary to maintain the refined grain size accomplished during working.

Leaded copper-zinc alloys are generally not recommended for forging applications. If these alloys are employed, special care must be taken in the design to completely support the structure in the die and eliminate deep-drawing applications. Lead is soluble in  $\beta$ -brass up to 2%, which allows hot working of alloys such as Cu-40%Zn brass. By contrast, 0.1% Pb in an  $\alpha$ -brass (Cu-30Zn) will result in hot cracking at forging temperatures. Insoluble lead can become liquid at forging temperatures, resulting in a noncompressible material in the forging and leading to catastrophic failures.

## ACKNOWLEDGMENT

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# Failures of Mechanical Springs

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MECHANICAL SPRINGS are used in mechanical components to exert force, provide flexibility, and absorb or store energy. Springs are made in a variety of shapes and sizes, ranging from delicate hairsprings for watches to massive buffer springs for railroad equipment. In general, springs may be classified as wire springs, flat springs, or special-shaped springs, with several variations within these divisions. The vast majority of springs are made from one of several grades of carbon or alloy steel. However, to accommodate the broad spectrum of applications, springs have been made from a wide variety of metals and alloys, including stainless steels, heat-resisting alloys, copper alloys, and titanium alloys. A comprehensive review of spring steel materials can be found in Ref 1.

The broad range of compositions, sizes, and types of springs is equaled by the required range of quality and characteristics. Quality can range from that required for low-cost ball-point pens to the extremely high quality necessary for aerospace and biomedical applications.

Ordinary springs, such as those used in non-corrosive environments and at temperatures not much higher than room temperature, are produced to high strength levels that often can be obtained entirely by severe cold working (drawing or rolling). This approach to producing springs generally dictates that no metal be removed from the wire or strip during working, which allows any existing defects in the raw material to remain. Springs produced in this manner are adequate for many applications because minor surface discontinuities are of little or no consequence.

More rigorous applications are likely to require springs with defect-free surfaces. Such surfaces can be obtained by conditioning of billets or even by grinding the wire or strip stock from which the springs are formed. These practices, however, significantly increase cost and decrease availability.

Additional information about stainless steel spring wire can be found in Ref 2.

## Operating Conditions

Although most spring applications involve operating temperatures that are not much

above or below room temperature, many applications require springs to operate over an appreciable temperature range. For instance, valve springs in internal-combustion engines must operate in frigid weather for start-up, then, within minutes, must function at the normal operating temperature of the spring chamber, which may approach 90 °C (194 °F). This temperature range, and even a range with a maximum temperature somewhat higher, can be accommodated by springs made of carbon or low-alloy steels. However, some springs are required to operate at much higher temperatures and thus must be made of heat-resisting materials. Such applications must be evaluated individually before a suitable selection of material can be made. While most of the failures discussed in this article involve fracture, it is also noted below that a spring will fail to perform its function if it relaxes due to temperature exposure beyond the limits of its material stability.

## Corrosive Environments

Most springs are not subjected to environments more corrosive than humid air. There are, however, many exceptions. As for elevated-temperature service, applications that involve corrosive environments must be evaluated individually before selection of materials for springs or of suitable protective coatings.

## Common Failure Mechanisms

More than half of all metallic springs that fail in testing or service do so by a fatigue mechanism. Many times, the fatigue crack is initiated by preexisting defects or cracks, corrosion, wear, or mechanical damage. However, in the majority of cases, there is no significant identifiable cause that initiated the fatigue failure. Once a fatigue crack has initiated, repeated applied stresses will cause the crack to incrementally grow until the cross-sectional area of the spring can no longer sustain the applied load, at which time overload failure will occur. A detailed discussion of the fatigue fracture mechanism, performance factors, residual stresses, surface

roughness, fatigue (*S-N*) curves, and design issues can be found in (Ref 3).

Relaxation is another common failure mechanism. This condition is caused by overstressing and often results from use of a grade of material that is marginal for the application. Relaxation from operation at excessively high temperatures is also a common failure mechanism, as noted above. For more information, see the article "Fracture Appearance and Mechanisms of Deformation and Fracture," in *Failure Analysis and Prevention*, Volume 11 of the *ASM Handbook*.

Although some spring materials are quite notch sensitive and have limited ductility, brittle fracture is not a major problem in springs in the usual sense. Apparently, this is true because the section sizes are usually small, an extreme degree of working has been used (which creates a homogeneous structure), and care in minimizing stress concentrations has been applied.

One form of brittle fracture that is occasionally encountered results from hydrogen absorption in martensitic structures, which occurs in electroplating and other cleaning and finishing processes. Corrosion contributes to spring failure in much the same way as it contributes to failure of other highly stressed structures, although in springs it is most often connected with fatigue. Nevertheless, static stress-corrosion failures can occur at high stresses in the presence of electrolytes.

## Examination of Spring Failures

For the most part, procedures that apply to analysis of failures of other parts also apply to analysis of spring failures. Loading and other service conditions as well as all details of the material should be determined, and a preliminary visual examination should be made to identify fracture surface features, wear patterns, contact marks, surface deposits, and temperature colors. Fracture features often are damaged by clashing of the portions of the broken spring. Therefore, it may be necessary to analyze several similar failures to arrive at the correct conclusion as to the cause of failure. For small springs, a stereoscopic bench microscope is essential because the entire cross

section may be only as large as the origin of failure in many large sections. As discussed in other sections of this volume, scanning electron microscopy (SEM) is an essential process for analyzing spring failures, particularly with energy-dispersive spectrometry (EDS), to confirm chemical identification of material impurities and surface damage.

## Common Causes of Failures

Overstressing is a common cause of spring failure. Settling and other types of distortion are generally caused by operation of springs at stresses that are higher than expected. Note that the stresses a given spring can withstand are greatly affected by the operating environment. For example, helical springs made of 6150 steel provided failure-free service in fuel-injection pumps when the fuel oil being pumped was a normal low-sulfur grade. However, several of these springs failed under identical stress conditions when the fuel oil contained substantial amounts of hydrogen sulfide. In addition, operating temperatures that are higher than those anticipated often result in failures of springs without changes in stress.

Design deficiencies, material defects, processing errors or deficiencies, and unusual operating conditions are common causes of spring failures. In most cases, these causes result in failure by fatigue. This article presents examples of specific spring failures. Although most of the springs discussed in these examples failed by fatigue, the fatigue cracks were initiated by a wide variety of causes. In some of the examples, more than one cause contributed to failure.

## Failures Caused by Material Defects

Seams frequently contribute to spring failures, usually by initiating fatigue cracks. Pipe solidification, inclusions, and internal voids may also contribute to premature failures. The following three examples describe fatigue failures that resulted from these types of material defects.

### Example 1: Fatigue Fracture of Alloy Steel Valve Springs

#### Background

Two outer valve springs broke during production engine testing and were submitted for laboratory analysis. The springs were from a current production lot and had been made from air-melted 6150 pretempered steel wire. The springs were 50 mm (2 in.) in outside diameter and 64 mm (2.5 in.) in free length, had five coils and squared-and-ground ends, and were made of 5.5-mm (0.2-in.)-diameter wire.

#### Investigation

The spring (Fig. 1a) broke approximately one turn from the end. Fracture was nucleated

by an apparent longitudinal subsurface defect. Magnetic-particle inspection did not reveal any additional cracks or defects along the length. Microscopic examination of transverse sections of the spring adjacent to the fracture surfaces revealed that the defect was a large pocket of nonmetallic inclusions at the origin of the fracture. The inclusions were alumina and silicate particles. Partial decarburization of the steel was evident at the periphery of the pocket of inclusions. The composition and appearance of the inclusions and the presence of the partial decarburization indicated that the inclusions could have been associated with ingot defects during the original steel-making solidification. The defect was 0.8 mm (0.03 in.) in diameter (max), 25 mm (1 in.) long, and 1.3 mm (0.05 in.) below the surface (Fig. 1b).

The steel had a hardness of 45 to 46 HRC (hardness Rockwell C), a microstructure of tempered martensite, and a grain size of ASTM 6 to 7, all of which were satisfactory for this application. The fracture surface contained beach marks typical of fatigue and were at a 45° angle to the wire axis, which indicated torsional stress to induce fracture.

#### Conclusion

The spring fractured by fatigue; fatigue cracking was nucleated at a subsurface defect that was longitudinal to the wire axis. The

stress concentration associated with the subsurface defect was responsible for the fracture (Ref 4).

### Example 2: Raw Material Defect in an Automotive Coil Spring

The existence of a raw material defect within the steel matrix, such as a nonmetallic inclusion, can have adverse effects on the fatigue performance of springs depending on the size and location of the defect and the application load of a coil spring (Ref 5). Figure 2(a) shows the location of a fatigue crack on the cross section of a spring that initiated approximately 1 mm subsurface due to a large inclusion particle. Figure 2(b) shows the fracture with SEM imaging. The EDS mapping scans in Fig. 3 reveal that the inclusion is rich in aluminum and magnesium. This particular coil spring failed prematurely at the location of the defect, despite all other parameters being normal.

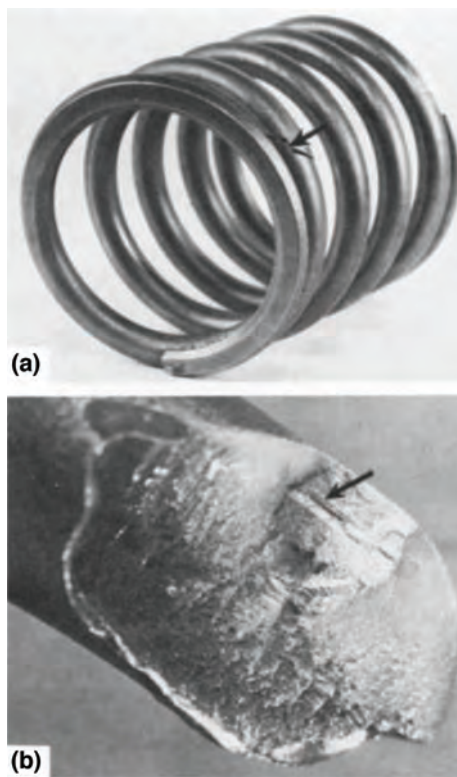
### Example 3: Preexisting Hot Mill Rolling Crack

Figure 4 shows a surface flaw deep enough to cause a coil spring to fail early; this surface imperfection was inherited from the raw material (Ref 5). This type of defect can occur when the surface flaw detector does not function normally. It is usually easy to determine if such a flaw was inherited from the raw material and is not due to coil manufacturing. A preexisting defect usually has surrounding decarburization after the raw material is heated during coil manufacturing, whereas a surface defect caused by coil manufacturing is often not accompanied by decarburization. The white areas along the length of the crack reveal a low-carbon microstructure (ferrite) that will etch white when compared with the normal microstructure surrounding the crack. This means that the open crack was exposed to high temperatures during the hot mill rolling operations for an extended time and resulted in a decarburization layer. This layer will have a lower strength, and the crack will act as a significant stress riser for fatigue initiation.

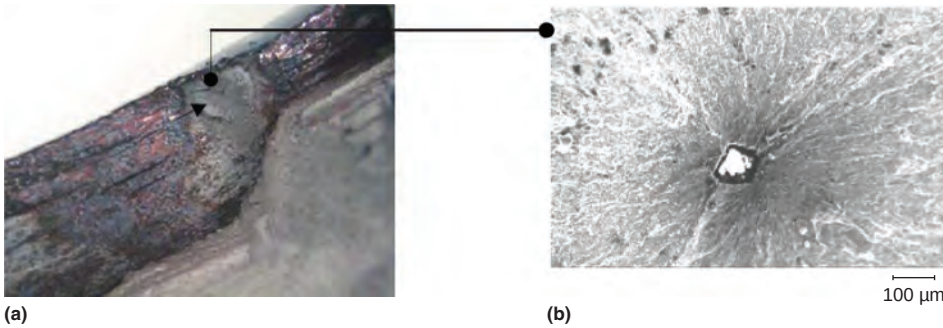
## Failures Caused by Fabrication

Fabrication errors or deficiencies often result in spring failures. The most common of these are:

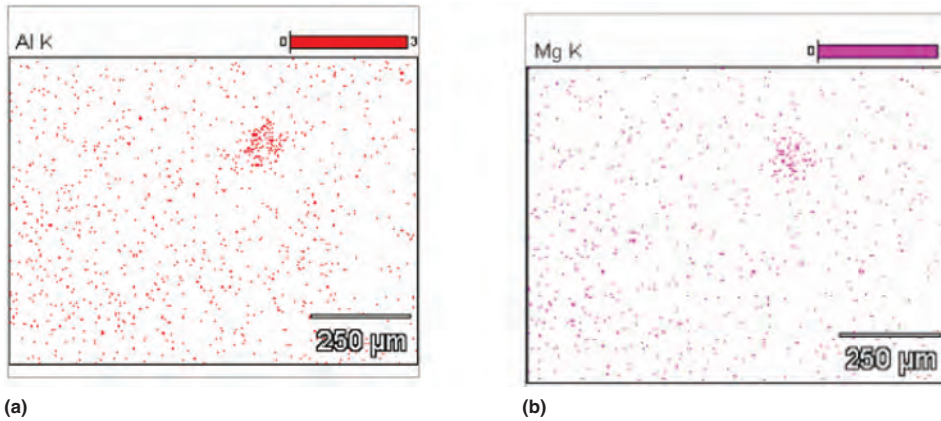
- Split wire, usually caused by overdrawing of the wire
- Tool marks
- Arc burns, sometimes occurring during welding and sometimes during plating; either can cause hard, brittle spots
- Hydrogen damage, usually resulting from improper pickling or plating practice
- Improper winding, which may allow formation of flats or scratches



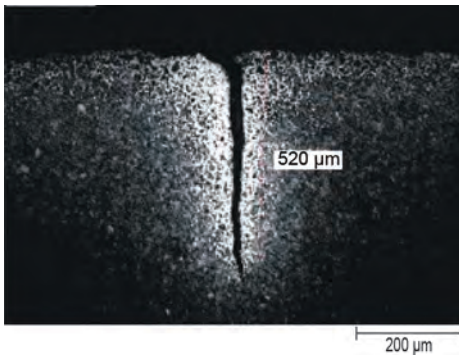
**Fig. 1** Valve-spring failure due to residual shrinkage during solidification. (a) Macrograph showing fracture, as indicated by arrow. (b) Fracture surface; pipe is indicated by arrow. Source: Ref 4



**Fig. 2** (a) Fatigue fracture surface of a coil spring that failed prematurely due to a subsurface inclusion; (b) as shown with SEM imaging



**Fig. 3** EDS results showing the inclusion is comprised of (a) aluminum and (b) magnesium. Scan was taken from the image in Fig. 2(b)



**Fig. 4** A preexisting hot mill rolling crack along the outside surface of the steel. Note the white indications of decarburization along the crack

- Improper heat treating
- Improper relation of spring index and hardness; a small index and high hardness may result in incipient cracks
- Unfavorable residual-stress pattern

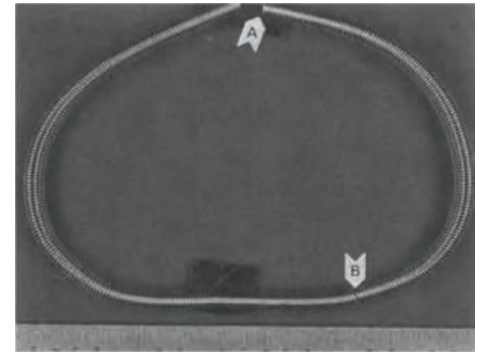
The examples that follow demonstrate how improper fabrication can result in premature fatigue failure.

#### Example 4: Failure of Stainless Steel Springs Used in an Oil Ring Lip Seal

##### Background

Two stainless steel springs in oil ring lip seals failed (Ref 6). Lip seals are used to seal both rotating and reciprocating shafts and consist of a polymer seal that contains an internal spring. The spring allows the seal to deform and remain tight against the shaft during rotation. The lip seals were used to seal a shaft between a turbine power unit and a pump in a commercial aircraft secondary unit, which began leaking oil heavily. Removal and inspection of the seals showed that the internal springs had fractured.

The springs were made from AISI type 302 stainless steel, work hardened by drawing to 2140 to 2240 MPa (310 to 325 ksi). The two ends of a single length of spring were wound together to make a loop used in the lip seal. Therefore, mating fracture faces were located on opposite ends of each broken spring (Fig. 5). Parts similar to these had been used successfully in this application for 16 years. These seals were from newly manufactured units, thus failing very early in the product life cycle.



**Fig. 5** Optical photograph of a failed spring showing the relation between the fracture (arrow A) and the joint (arrow B)

##### Investigation

Optical examination indicated flat (fatigue) fracture regions initiating on the inside of the coils, with different overload fracture morphologies. Although some details of the fracture surface could be examined optically, improved photographs were obtained via SEM. The fracture was located nearly opposite the joint where the spring was intertwined to form the loop; that is, the joint connection was at approximately the center of the broken spring length (Fig. 5).

The spring had a fatigue fracture that started at a drawing defect on the inside surface, as shown in Fig. 6, Fig. 7, and Fig. 8. The fracture was macroscopically flat (located nearly opposite the spring connection where the two ends were intertwined). The fatigue portion of the fracture was approximately one-half of the spring wire diameter (Fig. 6) indicated a moderate load application. The origin of the fracture was at a subsurface void that had a triangular-shaped cross section (Fig. 8). The void was associated with a wire-drawing defect, probably produced by a scratch or indentation in the die.

Metallographic analysis, chemical analysis, and hardness testing did not find any unusual results that may have contributed to the fatigue failure.

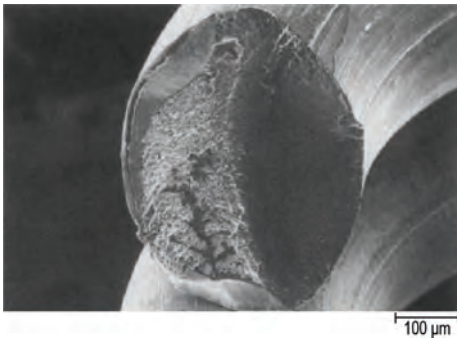
##### Conclusion

The results of the analysis showed that the springs failed by fatigue. The springs contained wire-drawing defects that were the origins of the fatigue cracks. Because this spring design had worked well in the past, the failures were attributed to these manufacturing defects.

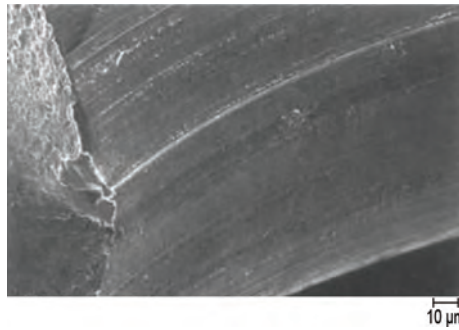
#### Example 5: Failure of Helical Compression Springs Due to Residual Stresses

##### Background

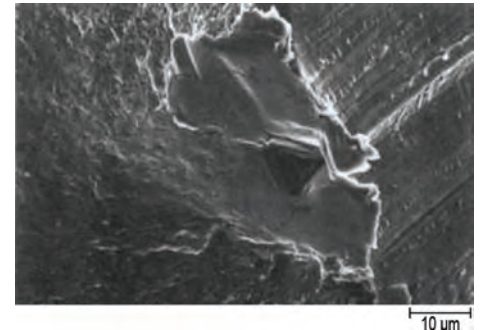
Two helical compression springs composed of chromium-silicon steel (50 to 54 HRC) failed early in service, although design and operating stresses appeared to be within the



**Fig. 6.** SEM image showing the fatigue crack on the right and overload area on the left



**Fig. 7** SEM image showing the drawing defect line at the initiation site of the fatigue crack in Fig. 6, magnification 400×

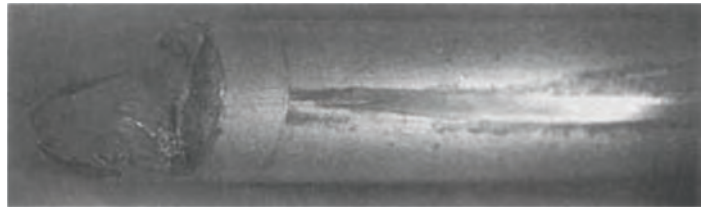


**Fig. 8** SEM image showing the initiation site in Fig. 7 at higher magnification. Magnification 1500×

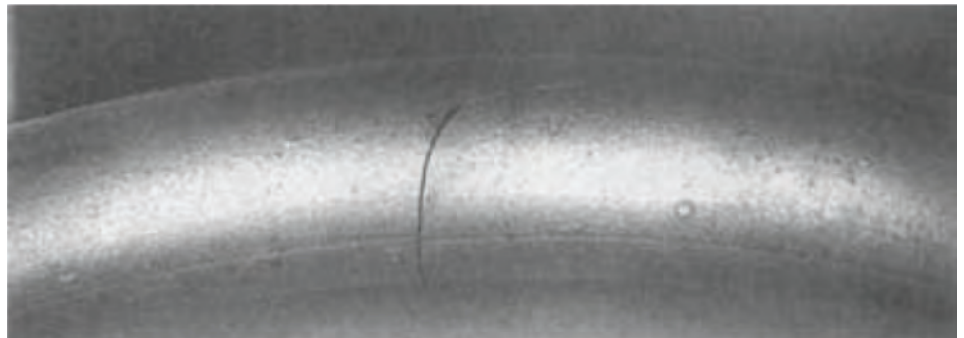
material endurance limit (Ref 7). Failure initiated from the spring inside coil surface; final fracture occurred in torsion. Helical compression springs are used in heavy truck engines. They may be subjected to corrosive environments, including exposure to exhaust gases. After approximately 80,000 km (50,000 mi) of service, broken springs were found in two engines. The trucks were used on public highways and were exposed to a variety of terrains and operating conditions, but conditions could not be characterized as extraordinary service. The springs were manufactured according to specification SAE J157, "Oil Tempered Chromium-Silicon Alloy Steel Wire and Springs." Shot peening and a surface coating were also required. The nominal wire diameter was 4.9 mm (0.192 in.).

#### Investigation

Figure 9 shows a helical spring that completely fractured, and Fig. 10 shows a second helical spring with a fatigue crack at a similar location as the fractured spring. A complete metallurgical analysis for microstructure and hardness did not find any significant issues. SEM imaging was performed (Fig. 11), and some evidence of corrosion was found; however, this was attributed to the harsh service environment and determined to not be a cause of the fatigue fracture. One of the fractured samples was evaluated with x-ray diffraction to determine the residual stresses along the inside diameter and outside diameter. The inside diameter had residual tensile stresses that ranged from 480 to 690 MPa (70 to 100 ksi), and the outside diameter had residual compressive stresses that ranged from -230 to -350 MPa (-33 to -50 ksi). If the springs had high residual stresses from the coiling operation, then the coil inside diameter would probably be in tension; the center of the spring, in a neutral condition; and the outside diameter, in compression. Residual tensile stress at the coil inside diameter could cause a crack to propagate perpendicular to the axis of the wire until the intact cross section was reduced enough for final propagation by



**Fig. 9** A fracture surface (left) on a helical spring



**Fig. 10** A fatigue crack on a second helical spring at a similar location as shown in Fig. 9

torsional overload. The presence of the residual tensile stress on the spring inside diameter could be due to insufficient stress relief after coiling the spring. Indeed, when one of the failed springs was examined for the presence of residual tensile stress, fairly high stresses were found. Failure of the springs could probably have been prevented by effective shot peening.

Although the spring specifications included requirements for shot peening, the presence of tensile residual stresses on the coil inside diameter indicates that shot peening, if performed, was not effective. As further confirmation of the absence of shot peening, the surface layer microstructure did not exhibit the deformation expected from shot peening. If the surface had been shot peened to provide compressive residual stresses, then the spring surface would not have been subject to fatigue cracking.



**Fig. 11** The fatigue crack area (top) and overload location (below) with some minor corrosion residue

Conclusion

Failure occurred because of the excess tensile residual stresses that remained after spring forming. Residual stress cracking may have been exacerbated by the presence of humidity or of chemicals within the exhaust gases. Effective shot peening would have eliminated the tensile residual stresses at the surface.

Example 6: Poorly Shot Peened Surfaces in an Automotive Coil Spring

Poorly shot peened surfaces can also be classified as surface imperfections (Ref 5). Figure 12 shows a comparison between two coils that failed at similar locations but possessed completely different fatigue lives. The image on the left side shows that the surface was poorly shot peened and therefore exhibited a shorter life. The image on the right side shows that the surface was shot peened sufficiently and therefore had a much longer life. Improper shot peening conditions can result in deep surface imperfections that will adversely affect the fatigue performance by imparting stress risers on the surface.

Example 7: Improper Heat Treatment of an Automotive Coil Spring

Improper heat treatment can be often overlooked since a temperature difference in heating might not relate directly to the hardness of the material (Ref 5). Extensive metallurgical evaluations are usually needed to identify this type of problem. Improper heat treatment can also result in the microstructure containing some ferrite/pearlite instead of the required 100% tempered martensite. This latter type of defect is easier to identify due to the clear difference in hardness where the ferrite/pearlite microstructure will likely have lower or insufficient hardness depending on the manufacturing requirements.

Figure 13 shows two coils of the same product with varying microstructures due to different heat treatment temperatures. This defect usually occurs when the heating system does not reach the proper temperature or when the parts do not reach the optimum temperature due to large furnace loads. Again, referring to Figure 13, the left-hand side coil has a much shorter lifetime than that of the right side.

Example 8: Decarburization of a Coil Spring

Decarburization is another issue that must be understood as a defect depending on the application (Ref 5). Partial decarburization can be permissible in some spring wire applications; however, this can be a significant issue in some industries, and a comprehensive understanding of the design and manufacturing requirements is needed. When a complete, full ferrite ring around the circumference is found, the wire is always subject to rejection. Figure 14 shows a coil spring failure due to an excessive decarburized layer.

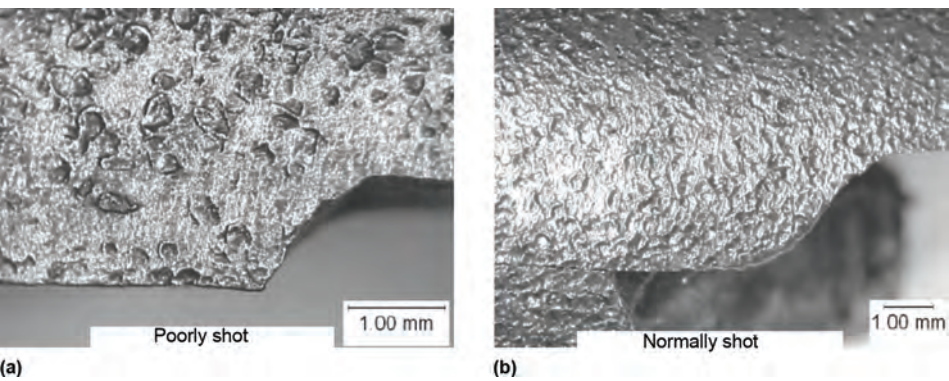


Fig. 12 Surface imperfections caused by (a) poor shot peening conditions that resulted in deep imperfections compared with (b) normal peening conditions

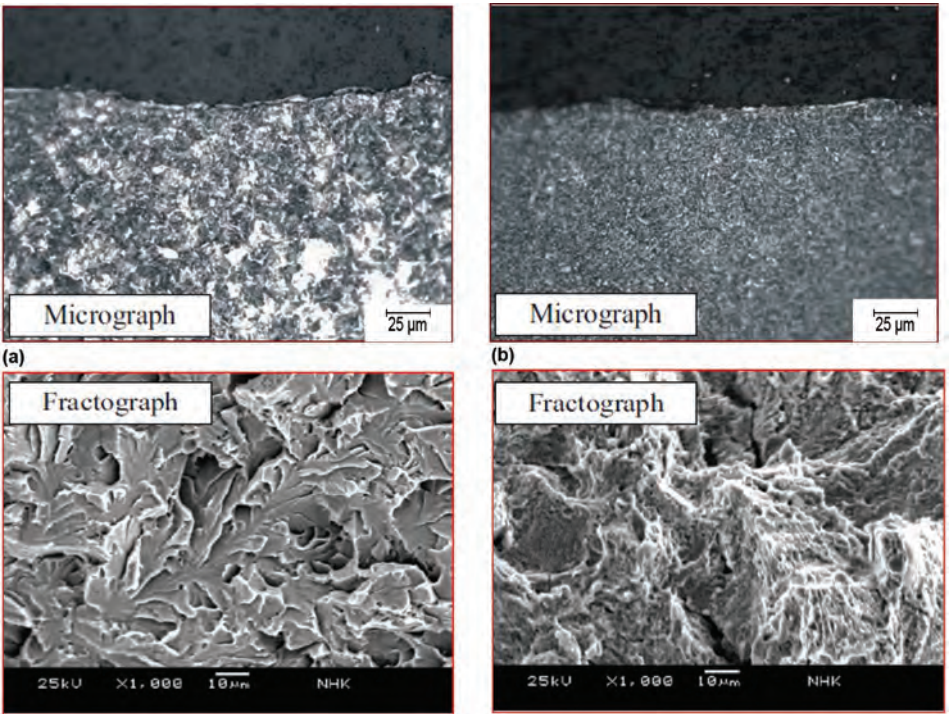


Fig. 13 (a) Improperly heat treated spring vs. (b) properly heat treated spring (tempered martensite) and the resulting fracture surface morphologies

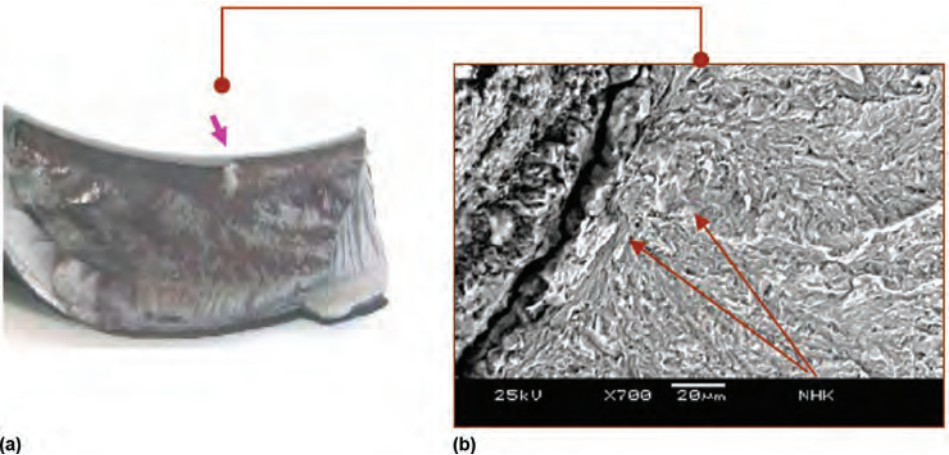


Fig. 14 (a) Fracture surface and (b) SEM fractography of a coil that failed due to excessive decarburization

The fractography of the sample also shows that the outer layer consisted of softer microstructures where almost no striations were visible. Further evaluation of the longitudinal cross section revealed that the sample had significant decarburization. Subsequently, when the microhardness near the surface was profiled, apparent decarburization was confirmed (Fig. 15).

### Failures Caused by Corrosion

Corrosive environments very often are principal contributors to failures of springs. Any metal is more likely to be attacked when under stress. Consequently, springs are extremely susceptible to stress-corrosion cracking and corrosion fatigue.

#### Example 9: Corrosion Fatigue Failure of a 302 Stainless Steel Spring

##### Background

After only two months of use in a printing operation, type 302 stainless steel springs were breaking into several pieces (Ref 3). The springs were operating over a very small deflection and were regulating the flow of ink, in which they were constantly immersed.

##### Investigation

Visual examination revealed clear evidence of fatigue fractures on every piece of the spring, and each of the fractures was oriented at 45° to the wire axis. Figure 16 shows crack arrest marks (beach marks), which are not always expected in the fracture surface of spring fatigue failures, as noted above. Clear evidence of pitting corrosion at the fatigue fracture origin can also be seen. However, the portions of spring showed no evidence of red rust and had a bright (greenish from the ink) surface, indicating that the corrosion pitting was local in nature. Analysis of the ink in which the spring was operating revealed that free-chloride ions were present. The extent of the pitting damage was surprising. Once initiated, the corrosion was rapid because the surface of the stainless steel could not passivate itself because there was no oxygen present in the alcohol-based ink.

##### Conclusion

The solution was to use an alternative ink that contained no free-chloride ions. Three years later, the springs were working satisfactorily. All the companies involved in this problem were surprised that corrosion fatigue was the problem because no rust could be seen and the operating stresses were insufficient to initiate fatigue.

#### Example 10: Stress-Corrosion Cracking of Inconel X-750 Springs

##### Background

Springs such as the one shown in Fig. 17 were used to tighten the interstage packing ring

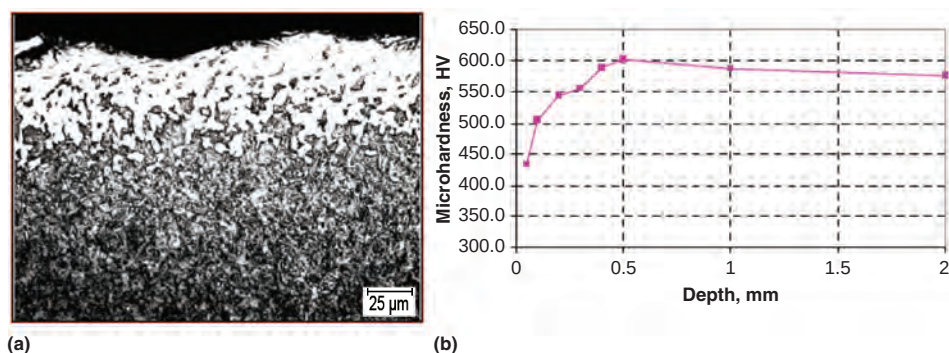


Fig. 15 Cross-sectional metallographic image of the (a) fractured coil and (b) microhardness profile near the surface

in a high-pressure turbine. After approximately seven years of operation, the turbine was opened for inspection, and several of the springs were found to be broken. Operation of the turbine was not impaired by the broken springs. The springs were made of 1.1-mm (0.045-in.)-diameter Inconel X-750 wire. The interstage packing rings were machined from centrifugally cast leaded nickel brass (German silver).

##### Investigation

There were chemical deposits in the turbine starting at about the fifth stage, and these deposits were judged to have come from the water or boiler chemicals. There were no broken springs in the fifth and sixth stages. The temperatures attained in the turbine stages were:

- Stage 1: 462 °C (864 °F)
- Stage 2: 405 °C (761 °F)
- Stage 3: 372 °C (702 °F)
- Stage 4: 351 °C (664 °F)
- Stage 5: 330 °C (626 °F)
- Stage 6: 313 °C (595 °F)

Many of the springs were broken into small pieces that had a white deposit and a rust-like surface scale. The white deposit was 100% water soluble and had a pH of 9.6 (slightly alkaline). Spectrographic examination showed that the surface scale contained a large amount of sodium; a small amount of tin; and traces of zinc, lead, and calcium.

Chemical analysis of the spring material showed that it was within specifications for Inconel X-750. The hardness values were within the specified range of 43 to 48 HRC.

Metallographic examination of a longitudinal section through the wire revealed intergranular cracks (Fig. 17) about 1.3 mm (0.05 in.) in depth and oriented at an angle of 45° to the axis of the wire. In some places, these cracks were 90° apart on the wire circumference. A light gray phase had penetrated the grains on the fracture surfaces. A fine corrosion residue was also observed along the surface of the wire to a depth of about 0.013 mm

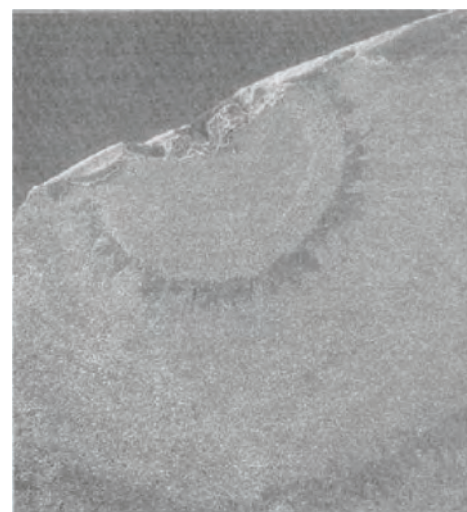


Fig. 16 Optical photograph of the fatigue region of a stainless steel compression spring that failed due to corrosion fatigue. Note beach marks on the fracture

(0.0005 in.). This general attack had the appearance of liquid-metal corrosion. When bent in the laboratory, the spring wires fractured in a brittle manner along a plane 45° to the wire axis, which is characteristic of brittle fracture from torsional loading.

High-nickel alloys are susceptible to embrittlement by some low-melting metals and alloys. Tin, lead, and zinc melt below the maximum service temperature of the turbine, which was 462 °C (864 °F). One tin-zinc system had a eutectic melting point of less than 204 °C (400 °F). Such metals can cause liquid-metal embrittlement at temperatures above their melting points. Sodium, tin, and zinc could have been present in a pigmented oil or grease used as a lubricant during spring winding and could have been left on the Inconel wire by the manufacturer. The white deposit on the surface of some of the springs was slightly alkaline and could have initiated stress-corrosion cracking by caustic penetration.

### Conclusion

The springs fractured by intergranular stress-corrosion cracking during service. The cracks were promoted by the action of liquid zinc and tin in combination with static and torsional stresses on the spring wire. Stress-corrosion cracking by caustic action was an alternative but less likely mechanism of fracture. Replacement springs were cleaned thoroughly by the spring manufacturer before shipment to remove all contaminants and were cleaned again before installation in the turbine. Failure of the springs was thus reduced to a minimum.

### Example 11: Aqueous Corrosion Fatigue of Type 302 and 304 Stainless Steel Springs

#### Background

After six months of operating a new chicken house, a farmer noticed that the majority of water feeders had stopped working (Ref 3). The water feeders operated on the principle that when the chickens pecked a plastic bowl, a compressed spring released a squirt of water. If the spring broke, no water came out.

#### Investigation

The small compression springs were made from 0.8-mm-diameter type 302 stainless steel, and the operating stresses were safely within the design limits given by the Goodman diagram. These springs were so small that it was necessary to use SEM to identify why they

had failed. Some were in three or four pieces, but most had only one fracture and did not show any evidence of mechanical or corrosion damage. On SEM evaluation, it was observed that the springs contained numerous cracks only on their inside surface, and these cracks were all at 45° to the wire axis. The cross patterns made by these cracks were recognized as the classic indication of corrosion fatigue.

#### Conclusion

The solution was to select a grade of spring steel that would be more corrosion resistant than 302 stainless. Glass bead peening would probably have been beneficial too, but the springs were too small for this process. Type 304 stainless steel springs were tried, but they also failed, albeit after two or three times longer service than those made of 302. Inconel 600 was eventually used, and no complaints about thirsty chickens have been received since.

### Example 12: Corrosion of an Automotive Coil Spring

Corrosion is a more common cause of automotive spring breakage than is usually understood by users (Ref 5). However, recent coating technology has reached a point where there is now a greater ability to withstand even the harshest stone impacts without chipping. Figure 18 shows the appearance of a coil spring that failed due to corrosion. The right

side of the figure depicts the cross section of the coil, with the red line illustrating its approximate original dimension. The material loss due to corrosion damage is apparent and will have an adverse effect on fatigue performance.

### Failures Caused by Errors in Design

#### Example 13: Analysis of Music Wire Springs Used in a Printer Mechanism

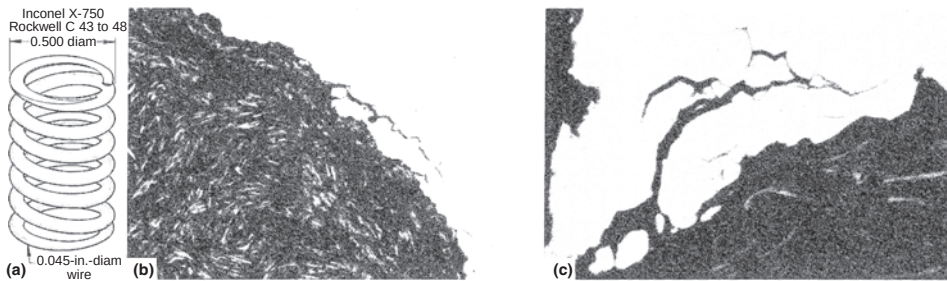
##### Background

Music wire springs used in a printer return mechanism failed near the bend in the hook portion of the spring during qualification testing (Ref 8). The music wire springs were specified for a new printer application. The springs were 21-coil music wire springs wound from 0.76-mm (0.030-in.)-diameter wire, which had been zinc plated (Fig. 19). The springs were manufactured from spring-quality music wire per ASTM A 228, which specifies an ultimate tensile strength of 2275 to 2515 MPa (330 to 365 ksi) for the 0.76-mm (0.030-in.)-diameter wire.

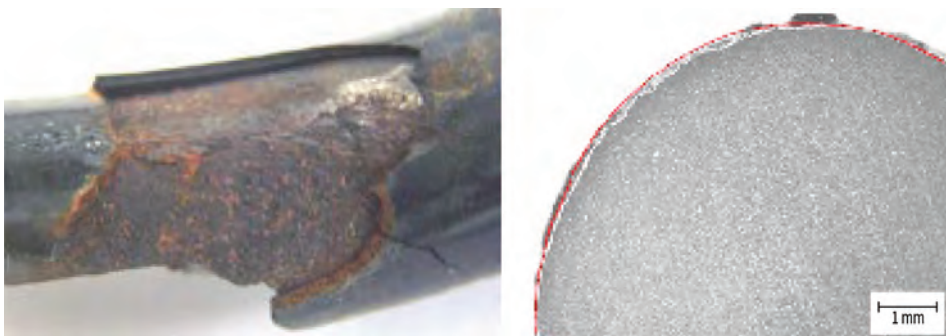
##### Investigation

All the springs had fractured in the hook portion (Fig. 19). Although the wire diameter was quite small, optical examination at 30× magnification indicated flat fracture regions initiating on the inside of the hooks, with several different overload fracture morphologies. Some details of the fracture surface could be examined optically, but much better photographs were obtained with SEM.

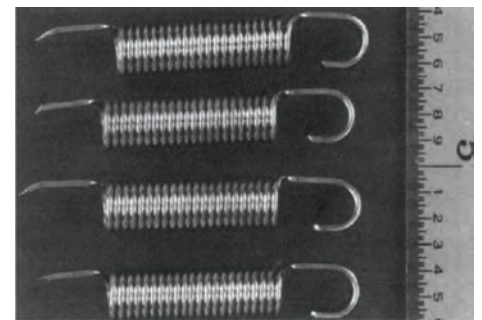
Figure 20 shows the typical failure geometry for all the springs examined. The fractures included a fatigue fracture, which was a flat portion on the inside of the bend extending approximately halfway through the wire thickness, and an overload fracture, which appeared intergranular toward the outside of the bend. In many of the springs examined, the inside of the hook had a rub mark on the surface at the origin of the fatigue fracture. The mark resulted from rubbing of the spring against the mating surface (Fig. 20). In one of the springs examined, there was no rub mark but



**Fig. 17** Inconel X-750 spring that failed by stress-corrosion cracking. (a) Configuration and dimensions (given in inches) of the spring. (b) and (c) Unetched longitudinal sections showing intergranular cracking. Magnification: 100× and 500×, respectively



**Fig. 18** Automotive coil spring that fractured due to corrosion



**Fig. 19** Optical photograph of several of the spring failures; the typical failure locations are shown on the left

rather a small, nonmetallic inclusion at the origin of the fatigue fracture (Fig. 21). The average fatigue striation spacing was approximately  $0.6\text{ }\mu\text{m}$  near the overload fracture, indicating that at least 500 load cycles were associated with the fatigue crack progression (Fig. 22). The overload portion of the fracture followed the elongated grains (a “woody” fracture) (Fig. 23) and should not be confused with intergranular fracture. Because the springs had been zinc plated, hydrogen embrittlement, which normally produces an intergranular fracture, was a specific concern. However, this did not occur; all of the spring fractures initiated in a fatigue mode.

EDS, metallography, and microhardness testing did not reveal anything unusual, and the microhardness results indicated the wire met the requirements of ASTM A228 when the values were converted to tensile strength estimates.

Fractography clearly established failure by fatigue as a result of the high loads caused by improper design of the springs. Improper design for fatigue is often encountered in qualification testing. In this analysis, it was important to rule out hydrogen embrittlement and material defects as the cause of failure in order to proceed with redesign. In these types of springs, the highest loads normally occur near the inside of the hook, and most spring fractures initiate in this area. Therefore, the fracture surface morphology near the inside of the hook is often the area of primary interest in spring failure analysis.

### Conclusion

The results showed that fatigue fractures in all the springs had initiated at rub marks on the inside surface of the hook, except for one spring, in which fracture had initiated at an inclusion on the inside surface of the hook. The rub marks were caused by rubbing of the spring hooks against the edge of a hole to which they were attached. This is a normal location for fatigue crack initiation in this type of coil spring because the highest tensile stresses are present on the inside surface of the hook. It was determined that the springs had been loaded beyond the fatigue strength (in bending) of the spring wire diameter used. A redesign using a spring with a larger wire diameter solved the problem.

### Example 14: Failure Analysis of Helical Suspension Springs Under Compressor Start/Stop Conditions

#### Background

During the start/stop testing of a compressor system, a noise developed that was unacceptable (Ref 9). During the investigation of the noise, it was found that one of the helical compressor springs was fractured. The failure occurred at 69K cycles, far short of the test goal of 500K. Subsequent to the first failure,

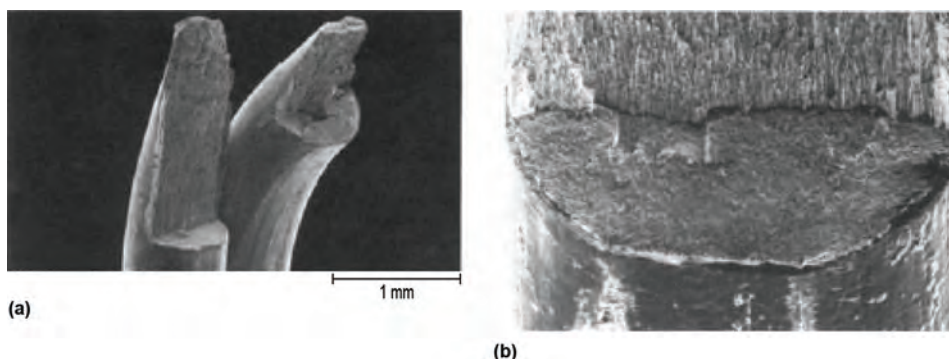


Fig. 20 SEM micrographs showing a rub mark at the fatigue origin, magnification 30×

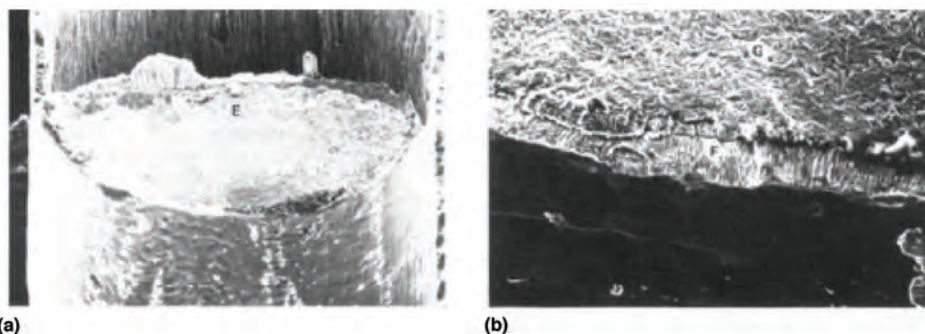


Fig. 21 SEM micrographs showing an inclusion at the origin

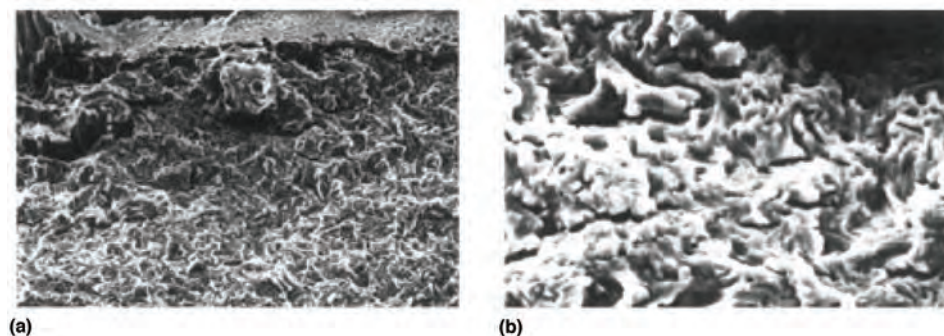


Fig. 22 SEM micrographs showing the fatigue portion of the fracture surface

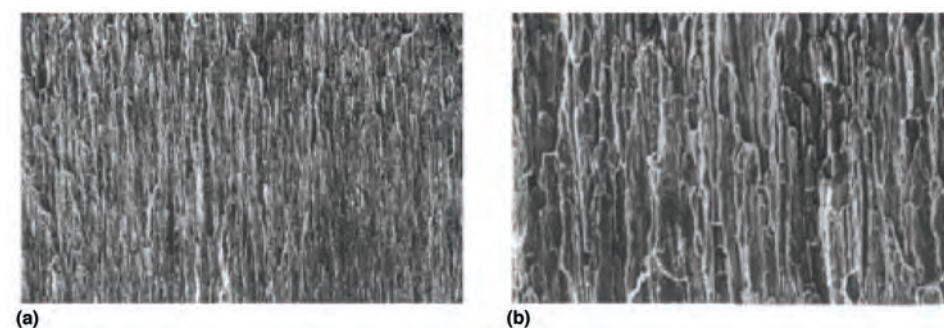
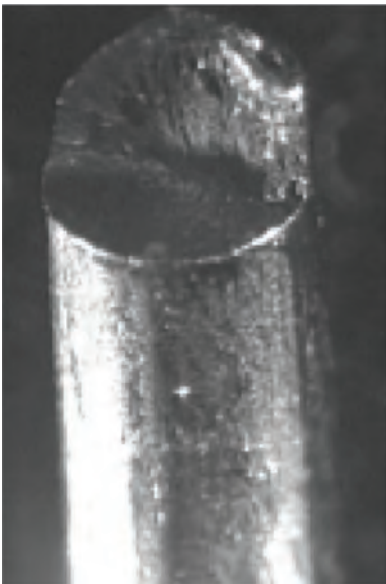


Fig. 23 SEM micrographs of the overload fracture showing a typical woody fracture appearance

all nine of the original units experienced spring failures. The helical springs were manufactured from ASTM 228 music wire with a diameter of  $1.5 \text{ mm} \pm 0.010$  and a specified tensile strength of 2025 to 2275 MPa (294 to 330 ksi). The actual music wire tensile strength was 2081 to 2165 MPa (302 to 314 ksi), within the specification for the wire.

#### Investigation

The spring was examined using SEM to determine the origin of the failures. Figure 24 shows the fracture surface of the spring, which occurred three active turns from the bottom. The fracture started near the inside diameter at a small angle just off the line perpendicular to the spring axis (Fig. 25). The fracture plane was near normal to the wire axis. The crack progressed across the diameter until it reached the approximate center of the section where it changed directions and ran parallel to the wire axis (material drawing direction) until the final fracture occurred in torsion. A tool mark near the origin due to the winding of the springs is also shown in Fig. 24 and 25, but the tool mark was not at the origin of failure. This type of tool mark can result in a stress concentration point along the inside diameter where the stresses are high. Due to the presence of this tool mark and the potential for fatigue crack initiation, the surface condition of the wire was considered marginal. However, since none of the springs showed indications of crack initiation at this location, it was determined not be the root cause of the failure. In general, the surface texture of these components appeared to be poor.



**Fig. 24** Photograph of helical compressor spring failure that occurred three active turns from the bottom. The failure origin was on the inside diameter of the spring

#### Conclusions

Fatigue analysis shows that the early failures were due to a poor surface texture on the springs. The poor surface texture was thought to be due to poor lubrication conditions during the drawing of the wire. The analysis of the compressor spring design from a fatigue standpoint showed that very little margin existed. The fatigue model can be used to investigate changes in the spring design to increase the margin in fatigue. Options that reduce the torsional shear stress for a given displacement can be explored. The current springs will not pass the reliability test based on the fatigue performance criteria when computer modeling is used. If the design of the spring cannot be changed, perhaps due to other test requirements, options should be explored to increase the fatigue performance of the spring. Shot peening the springs is one such option. Shot peening has been used to increase the fatigue strength of helical springs in many applications by inducing a compressive residual stress on the surface.

#### Failures of Shape Memory Alloy Springs

Shape memory or nitinol (NiTi) alloys have typical mechanical elastic properties and super-elastic properties that allow unique stress-strain relationships. Springlike components manufactured from a shape memory alloy can be used in medical treatments as they can be designed to be inserted into the body and then expand on reaching body temperature. These applications can result in unique spring failures.

#### Example 15: Nitinol Orthodontic Archwires

##### Investigation

The fracture mechanism exhibited by the NiTi wires was primarily ductile with a large

fraction of corroded areas (Fig. 26a, b, c, and d) (Ref 10). The large pitting corrosion regions were associated with superficial defects observed on the wires and acted as stress raisers to promote premature failure of the wires by overload. Microanalysis using EDS of the region adjacent to the corrosion surface (Fig. 26e) showed a composition of approximately 80% titanium and 20% nickel. These results were compared to the bulk composition of the alloy, revealing that the regions adjacent to the corrosion surface were depleted in nickel. This indicates the action of a mechanism of selective corrosion of nickel.

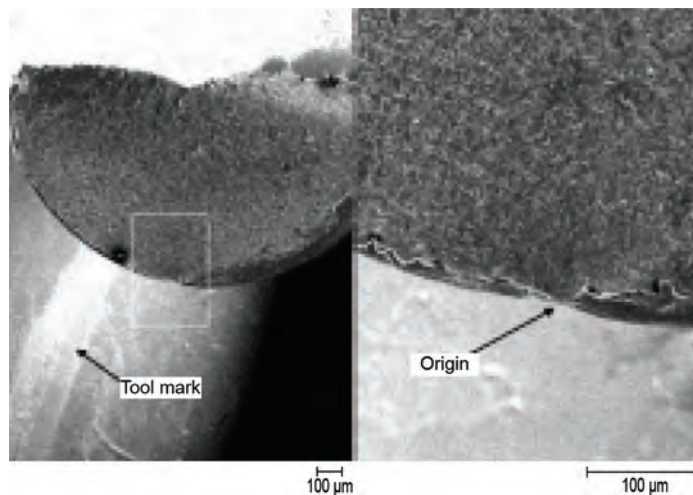
#### Conclusion

The selective attack occurred preferentially near the superficial defects that were generated during the manufacturing process (wire drawing and wire pickling). Previous work has shown that the presence of manufacturing defects dramatically decreases the corrosion resistance of NiTi alloys, because they remove the protective titanium dioxide layer. Additionally, NiTi shape memory alloys in the potentiostatic scratch tests present a difficult and slow titanium dioxide repassivation process, exposing the base metal to the selective dissolution of nickel and promoting pitting and crevice corrosion. Finally, the chemical compositions of the NiTi wires were not in accordance with ASTM specifications for wrought nickel-titanium shape memory alloys used in medical devices and surgical implants.

#### Example 16: Ultrasonic Cleaning-Induced Failures in Medical Devices

##### Investigation

A nitinol stent subjected to an unintentionally long ultrasonic cleaning treatment fractured during bend testing (Ref 11). SEM analysis showed that striations were present on the fracture surface, indicating that a



**Fig. 25** Scanning electron photograph showing the failure origin relative to the tool mark as a result of forming the spring. No internal material defect was found near the failure origin

fatigue crack preceded the overload fracture, as shown in Fig. 27 and 28. Several other cracks were observed at multiple locations along the stent (Fig. 29).

### Conclusion

The frequency of the ultrasonic cleaning excited resonance at the natural frequency of the stent and initiated a fatigue crack. At and near resonance, relatively small energy input

can result in large vibration amplitudes. These large deflection amplitudes at near-resonance conditions can and did result in fatigue crack initiation and growth.

### Example 17: Failure Analysis of NiTi Wires Used in Medical Applications

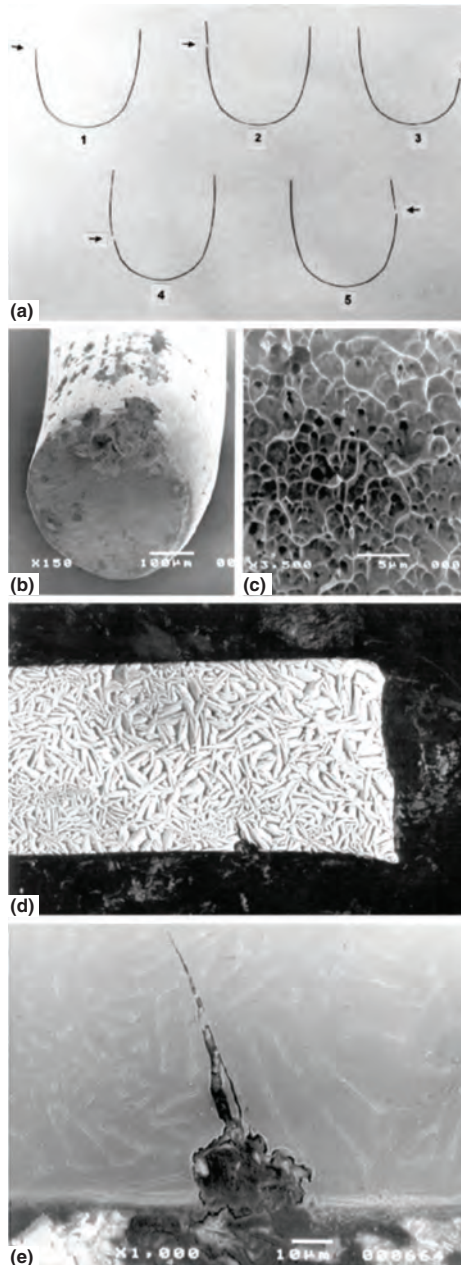
#### Investigation

A variety of fracture surfaces were produced under laboratory conditions for comparison in failure investigations (Ref 12). Results were generally as expected for a ductile material, with one exception. It was found that NiTi exhibits a unique damage mechanism under

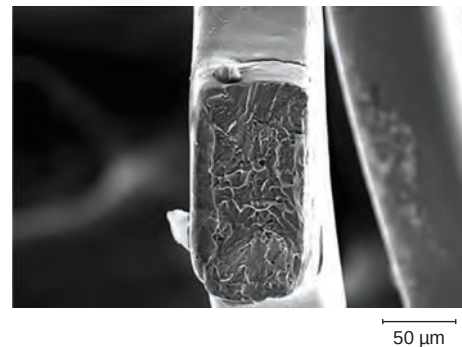
high bending strain, where damage occurs at the compression side of tight bends or kinks while the tensile side is unaffected. The damage begins as slip-line formation due to plastic deformation, which progresses to cracking at high strain levels. The cracks appear to initiate from slip lines and extend in a shear (mode II) manner.

As shown in Fig. 30, the compressive side of the bend surface was roughened and cracked, with no apparent damage on the tensile side of the bend. Progression of compressive damage at varying strains in one electropolished wire sample can be observed in Fig. 31 and 32. Compressive damage appears to begin as slip-line formation. The slip lines appear to deepen and eventually nucleate cracks (Fig. 31). Figures 32 and 33 show samples subjected to 50% and 31% compressive strain, respectively.

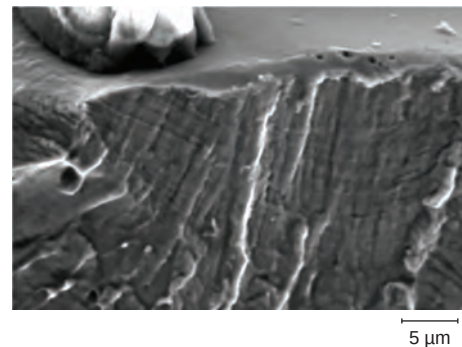
Consistent with cracking along slip lines, metallography showed that the cracks appear to propagate in compression in a shear/mode II manner (Fig. 34). Upon release of the bend, cracks were observed to extend further into the NiTi wire. This crack extension is believed to be driven by tensile stresses when the elastic strain energy of a severely bent NiTi wire is released.



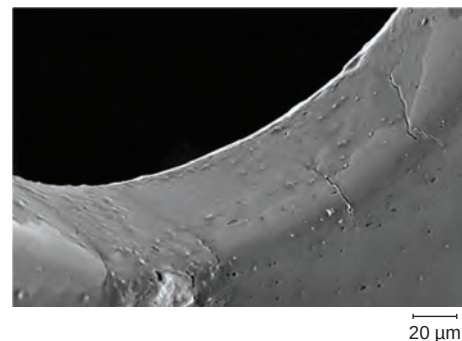
**Fig. 26** (a) Nitinol wires. (b) General view of the fracture surface showing surface defects and massive corrosion, magnification 150 $\times$ . (c) Detail of the fracture surface showing a ductile fracture, magnification 3500 $\times$ . (d) Acicular microstructure and the progression of lateral cracking. (e) Detail of the lateral cracking, magnification 1000 $\times$



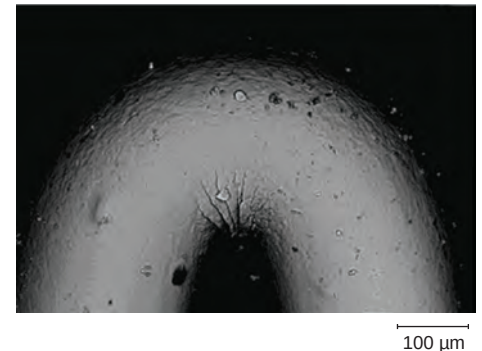
**Fig. 27** SEM image of fractured nitinol stent



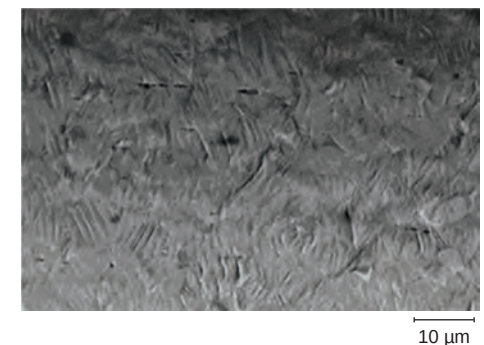
**Fig. 28** SEM image of striations on fractured nitinol stent



**Fig. 29** SEM image of cracks from other locations on the nitinol stent



**Fig. 30** SEM image showing a nitinol wire in compression with cracks at the intrados and no cracking at the extrados



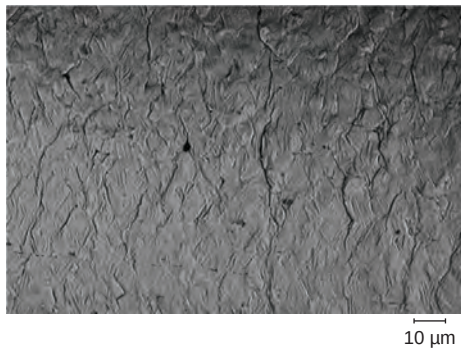
**Fig. 31** SEM image of slip lines and small cracks in a compression-damaged sample subjected to 27% strain

### Failures Caused by Operating Conditions

In some instances of spring failure, the cause is not evident, even after an exhaustive examination of material, processing, and design factors. When a thorough investigation in these areas does not reveal the cause, it is necessary to investigate operating conditions. Changes in operating conditions that are



**Fig. 32** SEM image of compression-induced cracks in a nitinol wire sample subjected to 50% strain



**Fig. 33** SEM image of slip lines and cracks in a compression-damaged nitinol wire sample subjected to 31% strain



**Fig. 34** Metallographic image of intrados cracks in compressed and released nitinol wire sample. Cracks initiate in shear during compression and grow in tension upon release

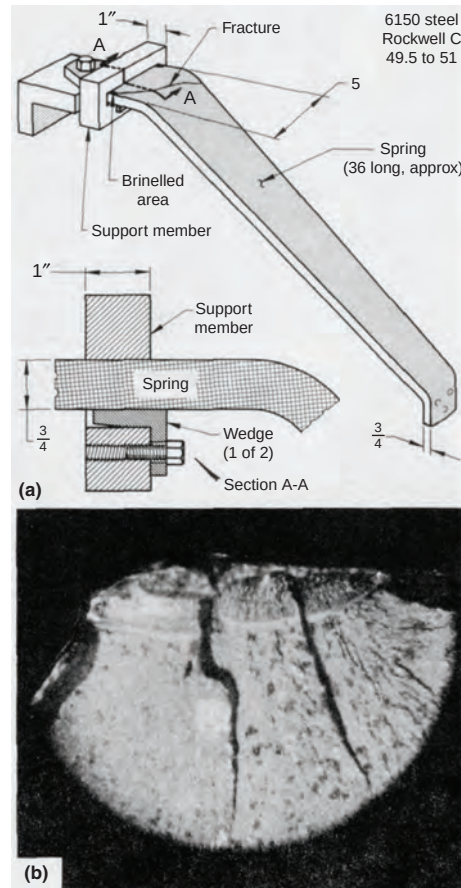
thought to be only minor deviations from normal sometimes prove to be the cause of spring failure.

### Example 18: Fracture of a Landing-Gear Flat Spring

A flat spring for the main landing gear of a light aircraft broke after safe execution of a hard landing. The spring failed (Fig. 35) was submitted for laboratory examination. No information was furnished concerning design, material, heat treatment, or time of service.

#### Investigation

The fracture occurred near the end of the spring that was inserted through a support member about 25 mm (1 in.) thick and attached to the fuselage by a single bolt. The required fit between the spring and the support member was obtained by inserting wedges beneath the spring at each side of the support. These wedges were bolted to the support member, and the degree of tightness of the assembly was controlled by the tension on the



**Fig. 35** Landing-gear spring, 6150 steel, that broke during a hard landing. (a) Configuration and dimensions (given in inches) of the spring. (b) Fractograph showing fatigue crack that initiated the brittle fracture; magnification, 7 $\times$

bolts. Details of the assembly are shown in section A-A in Fig. 35a.

The spring had broken laterally at an angle of about 60° to its longitudinal axis (Fig. 35a). One end of the fracture corresponded to the point at which the forward edge of the spring contacted the support member. The other end of the fracture was in an unsupported area.

Brinelling (plastic flow and indentation due to excessive localized contact pressure) had occurred on the upper surface of the spring where the forward and rear edges of the spring contacted the support member. There was no evidence of brinelling on the bottom surface of the spring where it contacted the support member.

Visual examination of the fracture surface revealed that the spring had failed by mainly brittle fracture. Chevron marks indicated that fracture had started beneath the brinelled area at the forward edge of the upper surface of the spring. Figure 35b, a light microscope fractograph of this corner, shows that the origin of the brittle fracture was in fact a small fatigue crack that had been present for a considerable period of time before final fracture occurred.

Microscopic examination of longitudinal and transverse sections near the fracture confirmed that the material had suffered extensive grain flow from cold working in the brinelled areas. The general microstructure was fine-grained, tempered martensite of good quality. There was no surface decarburization.

Chemical analysis identified the spring material as 6150 steel. The hardness of the material was 49.5 to 51 HRC.

#### Discussion

Normal recurring vertical landing loads cause the top surface of the spring at the edge of the support to be stressed in compression. These probably were the major loads applied to the spring. On the other hand, side and drag loads produce tensile stresses at the top surface of the spring. When vertical, side, and drag loads are applied simultaneously during a landing, the resultant stress at the support is compressive. However, after the aircraft has touched down and the vertical load has decreased to the static load of the aircraft, the continuing application of side and drag loads during taxiing causes a tensile stress at the top surface of the spring.

Compressive stresses in the top surface of the spring at the support are generally localized at the edges of the support by the wedges and by any lack of straightness across the section that may have been introduced during fabrication or that may have occurred when the complicated loading pattern caused a torsional deflection in the region of failure. Also, it was considered probable that the assembly could have been more tightly compressed than in the past. This would explain the brinelling observed near the forward and rear edges of

the top surface. The localized plastic deformation under compressive loading (brinelling) created high residual tensile stresses in the deformed region, reducing the capacity of the material to withstand fluctuating tensile stresses due to applied loads.

Ultimately, a crack started at the corner where the fluctuating tensile stress was highest under combined side and drag loads. This crack progressed by fatigue over a relatively long period of time (many landings). Final fracture occurred when the applied stress (primarily tensile stress due to side and drag loads) exceeded the critical fracture stress for the partially cracked spring. Under normal conditions, this type of cracked part might last a long time, although complete fracture would occur eventually. An abnormal condition, such as a sharp change of direction on a runway or taxiway, rolling over the edge of a runway, or encountering a rut, could impose a sufficient amount of additional loading to cause final, fast fracture.

### Conclusions

Fracture of the landing-gear spring was caused by a fatigue crack that resulted from excessive brinelling at the support point. The fatigue crack significantly lowered the resistance of the material to brittle fracture under critical side and drag loads.

Although the loading at the time of final fracture may have been unusual, a spring that did not contain a fatigue crack similar to the one found in the failed spring probably would not have broken. The fatigue crack had been present in the part for some time, and suitable inspection would have revealed it.

### Recommendations

Recommendations included regular visual examinations to detect evidence of brinelling and wear at the support in aircraft with this

configuration of landing-gear spring and means of attachment. If visual examination shows evidence of brinelling or wear, the spring should be magnetic-particle inspected for evidence of cracks. Liquid-penetrant inspection might be adequate if the amount of brinelling is minor, although care should be used in this type of inspection because brinelling can hide small cracks. Cracked parts should be replaced. Brinelling itself need not be cause for retiring a part. The presence of a crack was considered necessary for failure to occur in the manner described in this example.

### Example 19: Failure of a Retainer Spring Because of Cyclic Loading and Torsional Vibration.

The part shown in the "original design" portion of Fig. 36 was a valve-seat retainer spring from a fuel control on an aircraft engine. This spring was found to be broken during disassembly of the fuel control for a routine overhaul after 3980 hours of service. The two inner tabs had broken off but were not recovered. There had been no known malfunction of the fuel control.

The spring, made of 0.23-mm (0.009-in.)-thick 17-7 PH stainless steel, was thrust-loaded in assembly, but there was no thrust load on the tabs. The thrust load was the primary function of the spring. The radial, or torsional, load was rotational vibration of the assembly in which the spring imparted a relatively constant thrust load. The three tabs were antirotational devices and were subjected to unintentional torsional vibrations. These torsional vibrations created the cyclic loading that resulted in fracture.

### Investigation

As received at the laboratory, the spring was clean and free from corrosion products, with

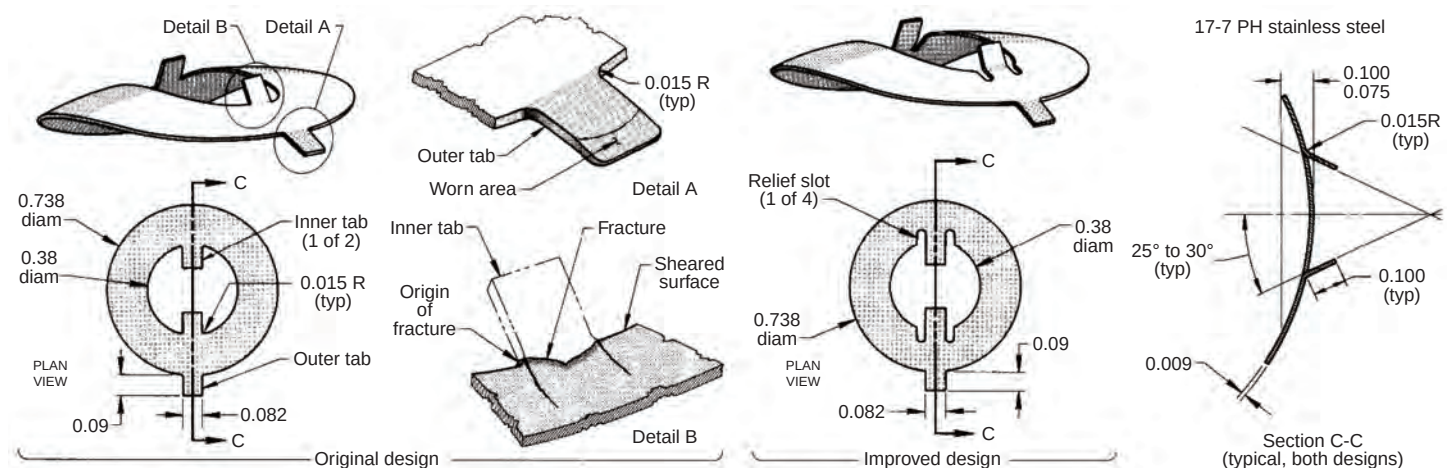
no evidence of chemical attack. Radial wear marks on the convex surface indicated that the part had been in relative rotation against its contacting member. The outer tab was worn in the area indicated in detail A in Fig. 36, which was evidence of slippage and apparent rotation against the contacting member. There was a wear mark on the concave surface 180° from the outer tab.

The fracture zone of an inner tab that broke off is shown in detail B in Fig. 36. Examination of the fracture surfaces of the washer in the area of the tabs revealed beach marks, indicating that fatigue fracture had been initiated at the convex surface of the washer and had propagated across to the concave surface. The cracks originated in the 0.38-mm (0.015-in.) radius fillet between the tab and the body of the washer. The fracture started at three places at radial cracks emanating from the radii at the intersections of the tab and washer and at one place on the sheared edge of the stamping.

Microscopic examination of the radial cracks showed a slight amount of plastic deformation, with the short surface between the fractures pushed toward the concave side. The two cracks on the sides of one tab were opened by sectioning. The fractures were smooth with arc-shape crack fronts.

Analysis of this compound fracture indicated that it was composed of fatigue fractures caused by the formed tab being loaded so as to compress the spring along the axis of its centerline and produce torsional vibrations. After the radial cracks had started as a result of torsional vibration, a straight bending mode of stress resulted in initiation of fatigue cracking on the concave side of the tab at the bend.

This was not an isolated failure. Shortly after this part had been examined, three additional springs that had broken were returned from service. Although there were some minor variations in the type of fracture, the basic cause appeared to be the same for all



**Fig. 36** Original and improved designs of a 17-7 PH stainless steel valve-seat retainer spring. As originally designed, the inner tabs on the spring broke off as a result of fatigue, and the outer tab exhibited wear

springs. Examination of a mating part revealed worn impressions that had been generated by the torsional motion of the tabs against the walls of the mating slot. Twelve springs from stock were inspected and found to be in satisfactory condition except for some malformed areas that were unrelated to this problem.

#### Conclusion

The two inner tabs broke in fatigue as the result of cyclic loading that compressed and torsionally vibrated the spring. The fracture resulted in relative rotation between the spring and the mating parts, both the one in contact with the convex side and the one in contact with the outer tab, as evidenced by the wear on the appropriate surface.

#### Corrective Measures

The springs were redesigned as shown in the “improved design” view in Fig. 36. The fillets were replaced with slots to minimize stress concentration at the corners. Fatigue testing indicated that springs of the improved design would have a service life 212 times as long as that of springs of the original design.

#### ACKNOWLEDGMENTS

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# Failure Analysis of Railroad Components

Hans Iwand and Joel Hassebrock, Engineering Systems Inc. (ESi)

## Freight Railroad Industry

The railroad industry is one of the most arduous engineering environments. While the rails, wheels, and axles continue to be manufactured from plain carbon steels, there are variations encountered in rail and wheel steels wherein microalloying and pearlite refinement are employed to increase the wear resistance of the steel. In addition, many of the components have long calendar lives. While most rolling stock is limited to 40 years of age, it is not uncommon to encounter components that are older, especially in lightly used rail.

Because of the tough environment, a primary mode of failure is fatigue. It is with good reason that freight car axles mimic the R.R. Moore rotating-fatigue sample geometry with generous radii and large cross sections. As the industry has become more competitive, loads have been increased, reducing the factor of safety with respect to fatigue life. Therefore, the existence of corrosion pitting and manufacturing defects, both surface and subsurface, has become more important.

As in all failure investigations, caution should be exercised by the analyst to gather as much background information as possible. Frequently, the ability to identify that a component failed due to fatigue failure is only partially adequate; the question of why fatigue fracture initiated at the specific location or component needs to be determined. For example, a rail can be seen to have failed due to a gage corner-initiated fatigue crack, but the rail was just evaluated by a track-based ultrasonic flaw-detection machine. Careful inspection of the wheel-rail interface reveals a pronounced network of surface cracking due to rolling-contact fatigue, that is, shelling. The shelled surface attenuates the ultrasonic sound, thus preventing a good reflection. In addition, the analyst is frequently required to decide if the failure was then also causative of a derailment. It is not uncommon to identify fracture surfaces containing signs of subsurface or surface-initiated fatigue morphology, but the fracture was caused by the derailling train and did not initiate or cause the derailment.

An additional example is the freight car wheel. Some wheels are forged, others are

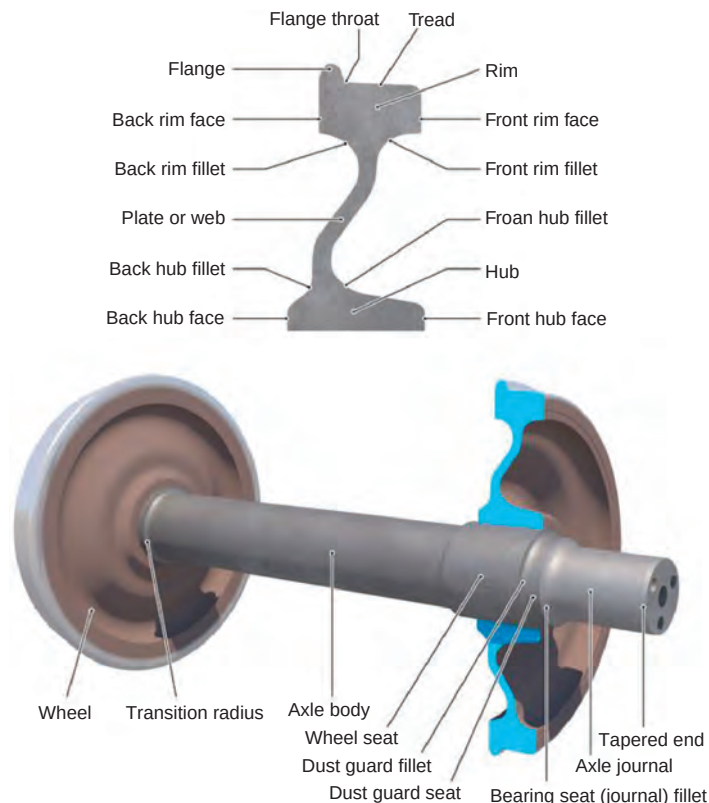
cast. North American railroads use primarily “single” wear wheels. However, the wheels are reprofiled when the wheelsets are reconditioned. The reduced tread thickness increases the Hertzian subsurface stress state, should inclusions be present. Inclusions that are innocuous when the wheel tread was full become more significant as tread thickness continues to decrease due to wear or reconditioning.

Electronics have become more compact and are heavily utilized in many of the locomotives, track geometry, ultrasonic rail test vehicles, track signals, and so on. The designer of such components must be cognizant of extreme temperatures, high dust (containing high iron-particle content), in addition to wide-ranging moisture conditions.

Many components of the railroad industry are covered elsewhere (engine components, lifting and rigging equipment, electrical, etc.) and, while encountered in the railroad industry, are not unique to the railroad industry. The following case histories are intended as an overview of the unique types of failures encountered in the freight railroad industry.

## Railroad Wheelset and Bearing Nomenclature

A wheelset is defined as two wheels, two cartridge-style tapered roller bearings, and an axle. The arrangement is shown in Fig. 1. Note that there are no bearings applied to



**Fig. 1** Wheelset nomenclature

the journals in Fig. 1; bearing nomenclature is shown in Fig. 2.

The manufacturing process of building the wheelset components is performed by axle, bearing, and wheel manufacturers. The process of manufacturing the wheelsets by pressing the wheels and bearings to the axle, as well as installing the end caps, locking plates, and cap screws, is performed by the wheel shop. Locking plate information located on the bearings indicates the wheel shop where the bearings were applied, the month and year of application, as well as the bearing reconditioner.

In some cases, all the wheelset components are new, that is, having the same or similar manufacture dates. In other cases, newly manufactured wheelset components are assembled with reconditioned components. Regardless of whether the subcomponents are new or reconditioned, the wheel shop employs a variety of cleaning, inspection, and magnetic-particle testing techniques during assembly of the components onto the axle. A press is required to push the wheels and the bearings onto the axle,

because the inside diameter of the wheel hubs and bearing components is smaller than the diameter of the wheel seat and journal. This type of joint is referred to as an interference fit.

The dimensional tolerances provided in the Association of American Railroads (AAR) *Manual of Standards and Recommended Practices* (MSRP), Volumes H and H-II, *Roller Bearing Manual*, as well as Volumes G and G-II, *Wheel and Axle Manual*, ensure that the fit between the bore of the bearing assembly and the axle journal is tight enough but not so tight that too much interference fit is present during assembly.

## Axle Failures

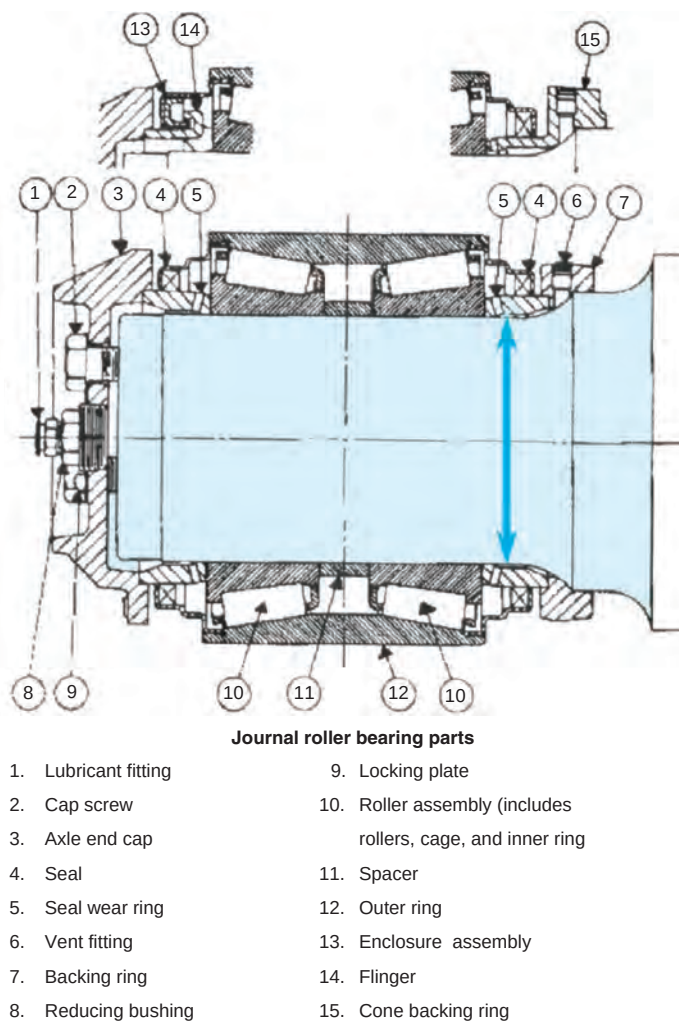
Axle failures in the rail industries have common failure locations—either at the journal fillet or in the barrel of the axle. (Case studies of common failure modes are provided in this article.) The latter often occurs from mechanical damage to the axle body from handling.

The mechanical damage leads to a stress concentration, which initiates rotational bending-fatigue failure. To prevent mechanical damage, special lifting attachments are used to handle the axles. Axles are not allowed to be weld repaired but can be affected by weld splatter or torch cutting of nearby parts of the railcar. Close observation of the work performed can prevent these failures.

Other less common failures of axles between the wheels are due to manufacturing defects such as forging bursts, which are large voids present within the axle. These should be detected by nondestructive testing inspection techniques; if not, they will cause failure due to rotational bending fatigue that initiates at a location within the axle and progresses outward.

## Axle Journal Failures Caused by Derailments

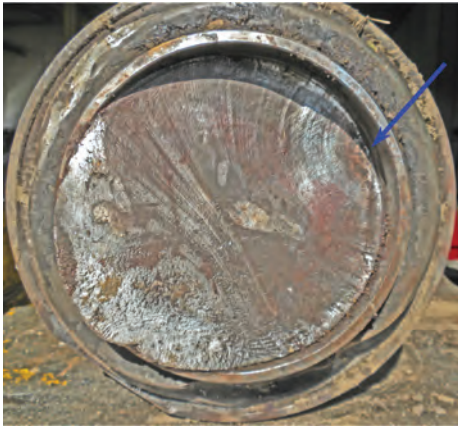
Two derailments occurred in close temporal proximity to each other, both caused by journal failures. The fractured journals from both derailments were due to progressive fatigue crack propagation (Fig. 3–6). The progressive fractures grew until the applied stress exceeded the strength of the remaining axle, and the journal fractured from the wheelset. By following the beach marks to the point of origin, it is possible to locate the fracture-initiation site, indicated by the longer arrow in Fig. 3 from derailment 1 and the two arrows in Fig. 6 from derailment 2. After removal of the bearing from the journal, it was apparent that the journal surface had a pronounced line of galling. In addition, the galling aligned with the fracture-initiation sites being the fatigue cracks for both derailments.



**Fig. 2** Tapered roller bearing nomenclature. Note: The blue arrow indicates the inner seal wear ring position; the journal is colored blue for clarity. Source: Ref 1



**Fig. 3** Initiation site of fracture surface is indicated by larger blue arrow. Beginning of galling damage is indicated by smaller blue arrow



**Fig. 4** Initiation site of axle 1, indicated by blue arrow, bearing side of journal



**Fig. 5** Axle 1 wheel and backing ring in place. Arrow indicates fracture-initiation site



**Fig. 6** Derailment 2. Arrows indicate boundary of galling damage and fracture-initiation location

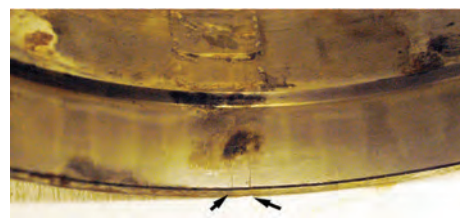
Each axle stub was pressed out of the bearing bores, and samples were sectioned from the fracture surface at and away from the initiation site. Mechanical property samples were also removed.

After disassembly, it was noted that the surface of the journals had been heavily scored, and the inside of the inner seal wear ring contained unique wear patterns (Fig. 7, 8). In one case, the seal wear ring had cracked (Fig. 8).

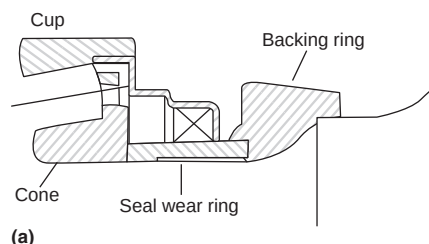
According to Ref 2, "Galling can be considered a severe form of adhesive wear. With high loads and poor lubrication, surface damage can occur on sliding metal components. The damage is characterized by localized macroscopic material transfer, that is, large fragments or surface protrusions that are easily visible on either or both surfaces. This gross damage is usually referred to as galling, and it can occur after just a few cycles of movement between the mating surfaces. Severe galling can result in seizure of the metal surfaces." The AAR MSRP requires that a coating of castor oil or heavy mineral oil be applied to the surface of the axle journal prior to mounting



**Fig. 7** Black arrows indicate galling damage. Note the fretting discoloration at right and left



**Fig. 8** Cracking present on seal wear ring due to excessive hoop stress



**Fig. 9** Schematic showing increased contact area in the updated seal wear ring design. Note the contact area of the seal wear ring in (a) the conventional design compared to (b) the updated design

the bearing, to provide lubrication during the press operation (Ref 3).

There are several factors that can affect the likelihood that galling will occur during an assembly operation. In this case, referring to Fig. 2, it can be seen that there are four components within the bearing that have an interference fit to the axle journal (in order of assembly):

- Inner seal wear ring
- Inner roller bearing race
- Outer roller bearing race
- Outer seal wear ring

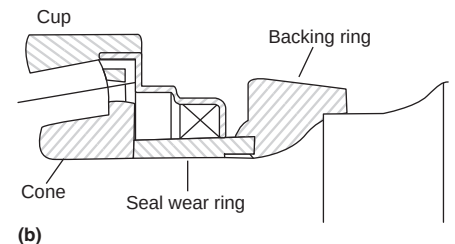
As each component is pressed over the lead-in taper of the axle journal, additional force is required. Once the bearing is completely pressed onto the axle, a bearing seating force of  $50 \pm 5$  tons (F-size bearing) must be achieved to ensure that the bearing is properly seated (Ref 4).

Because the bearing/wheelset design is a cantilever arrangement, there are bending forces applied to the end of the axle. These bending forces cause the axle journal, and therefore the bearing components, to rub against the contact surfaces. To address the fretting that occurred between the inner seal wear ring and the axle journal, a uniquely designed seal wear ring was introduced to the market.

At the time of introduction, there were several iterations of the lead angle of the seal wear ring. The lead-in angle is the angle of the chamfer between the larger inside diameter of the seal wear ring and the smaller inside diameter of the seal wear ring. The initial angle was  $30^\circ$ ; approximately eight years later, it was shown to be nominally  $12^\circ$ .

The significance of the lead angle and the increased contact area between the axle journal and the inner seal wear ring is that both design characteristics also influence the likelihood that galling can occur during bearing installation. The design changes are noted in Fig. 9. The design characteristics of the seal wear ring are by no means the only variables that can cause galling to occur; a variety of variables exists (Ref 2), such as (the following list should not be considered exhaustive):

- Rate of bearing to axle journal application
- Cleanliness
- Alignment of press and bearing components



- Amount and quality of lubrication present
- Relative temperature of components
- Surface finish of journal
- Surface finish of seal wear rings and bearing press surfaces

The occurrence of journal galling from bearing installation does not always cause a derailment to occur. While having all of the variables recorded is an ideal situation, it is still possible to use deductive reasoning and engineering insight to ascertain the most likely cause of galling, specifically, the amount of galling damage encountered by axle 1 and then axle 2.

Wheel shops would normally mount bearings from various bearing manufacturers, depending on their client's specifications. Were the galling inherent in a specific wheel shop and not others, one could perhaps conclude that the problem was localized to their operations. In this case, galling during installation of bearings containing the uniquely designed seal wear ring occurred in multiple wheel shops. Therefore, the common denominator among several wheel shops was the unique seal wear ring.

Interchange agreements between railroads are governed by the AAR Interchange Rules, which rely on the AAR MSRP; thus, all wheelsets that operate in railroad interchange must adhere to assembly rules as outlined in the AAR MSRP. The manuals define the requirements for successful and safe manufacture of any variation of new and reconditioned bearing/axle combinations. During the pressing of a bearing to an axle journal, a slip-stick phenomenon exists wherein the press reaches a force, the bearing rapidly slips on the journal, the press builds up force, and bearing slip again occurs. However, it remains the responsibility of the wheel shop to discern between typical manufacture variation and manufacture of defective product.

## Discussion

The two derailments occurred due to a failed axle, specifically, a failed journal. As part of the wheelset manufacturing process, both axle journals had been damaged during the pressing of the bearings to axle journals. Both axles failed due to progressive crack growth, referred to as fatigue fracture. The damage on both journals is defined as galling, which is an advanced form of adhesive wear. The galling damage was determined to be the initiation site from which progressive fatigue fracture propagated. These wheelsets were manufactured in a defective manner.

The AAR MSRP is recognized by the railroad industry as the governing rules for the manufacture of wheelsets and wheelset components. Allowing wheelsets to be entered into interchange service with galling damage such as was observed in the two axles was a violation of the AAR MSRP, G-II, Rules 1.2.4,

1.2.7.3, and 1.1.11. The "last line of defense" during wheelset manufacturing lies with the wheel shop, because there is no method of inspection once the wheelsets enter into interchange service. Once the wheelsets were placed into rail service, it was only a matter of time before the journals would fail and cause the derailments, unless the wheelset was removed for other reasons.

## Axle Journal Failure Caused by Corrosion Pitting

Broken journal failure mechanisms have been a recurring failure mode in the railroad industry. Because the bearing components cover the axle, the presence of cracks cannot be identified during railroad operations. History has shown that corrosion under the backing ring, within the fillet, can initiate fatigue cracking that leads to failure and subsequent derailment. The implementation of fitting backing rings has improved this issue, but failures persist, whether in new or reconditioned wheelset assemblies. It often has been found that failures in reconditioned wheelset assemblies occur within five years or less of reconditioning. The following is a case study of one such event, followed by a case study of failure from an original application that had journal failure after 30 years.

## Introduction

In this case, the locking plate stampings present on the failed journal indicate where and when the bearing had been reconditioned: March 2015 by a wheel shop. The derailment

occurred in June 2019. Figures 10 and 11 show sections of the wheelset prior to inspection.

## Wheelset Assembly

When building a wheelset, there are two methods that comply with the AAR MSRP. One method uses new components to build a wheelset, while the other uses reconditioned components, where needed, to make a "like new" replacement.

Used axles accumulate fretting debris and oil on the fillet and journal of the axle. (The fillet is the radiused surface blending the cylindrical journal to the dust-guard diameter of the axle.) Reconditioned axle journals are cleaned of debris, typically using light abrasives; in



Fig. 10 As-received axle stub with bearing in place



Fig. 11 As-received axle showing fracture surface

accordance with the MSRP, corrosion pits are to be removed within the fillet area.

The fillet and dust-guard area (just inboard of the fillet) are coated with Tectyl (Ashland Consumer Markets) product (Ref 5). Tectyl dries to a waxy consistency that, when properly applied, forms a sealing surface between the backing ring and the axle. To protect the sliding surfaces during application of the bearing and to provide a seal on the inboard side of the bearing, one of several oil-based materials is applied to the fillet and journal. A breakdown of the wheelset components is shown in Fig. 12.

Since the adoption of the 286,000 lb gross rail load (GRL) freight cars, all wheelsets must be assembled with fitted backing rings on fitted axles. This means that the backing ring must have an interference press-fit onto the axle. Figure 13 compares a fitted backing ring with a nonfitted backing ring arrangement. This axle was found to have a fitted backing ring. (For non-286,000 lb GRL axles, it is possible that fitted backing rings could be applied to nonfitted axles. The backing ring would cover the dust guard of the axle; however, the axle dust guard would be too small in diameter to provide a press fit.)

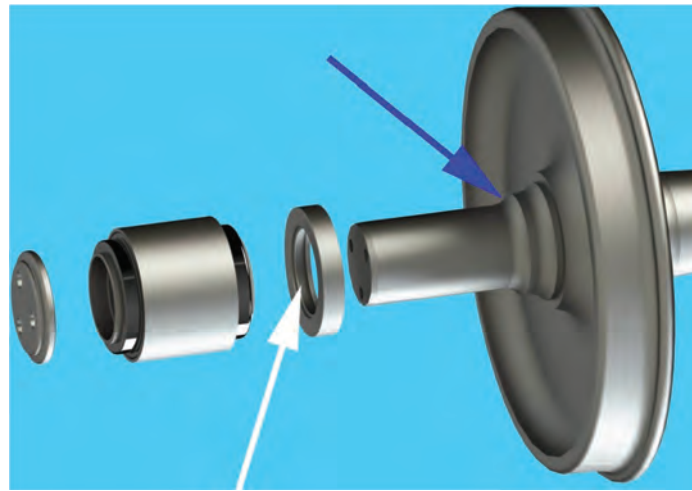
### Background

There is no routine maintenance interval (mileage or time) for bearings, wheels, or axles on freight car wheelsets. Maintenance rules are outlined in the AAR *Field Manual of Interchange Rules*, which includes Why Made (WM) codes that dictate wheel, axle, and bearing conditions that, if met, are cause for the wheelset to be removed from operation. With the exception of overheated, leaking grease, and noisy bearings, all of the other WM codes are directed to the wheels and axle condition. When a wheelset is returned to a wheel shop, the bearings must be removed and reconditioned prior to reapplication of an axle. Similarly, axles do not have an aging-out rule; as long as the dimensions meet the requirements as outlined in AAR MSRP G and G-II, the axles can be reconditioned and placed back into service in an “as-new” condition indefinitely. The same can be said of wheels. However, because the treads of the wheels do wear, it is generally true that when wheels are pressed off an axle, they will no longer be remounted to an axle.

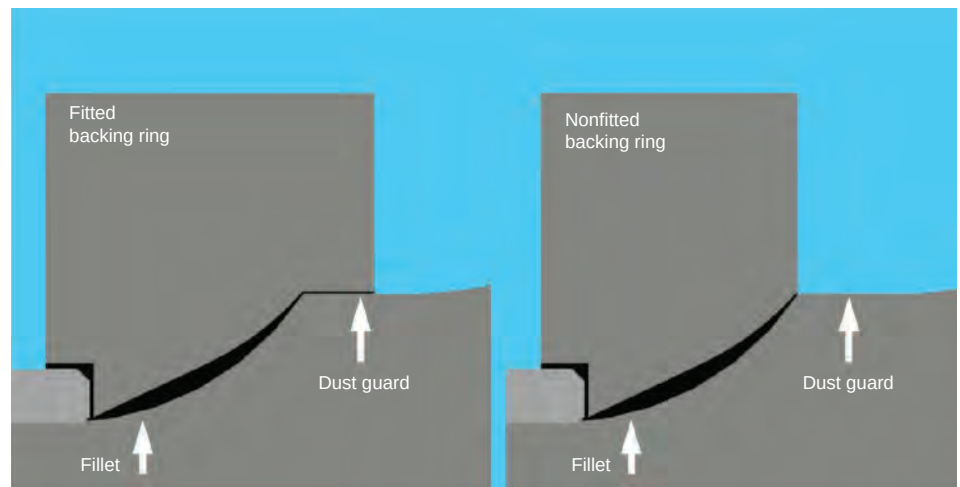
### Analysis

#### Axle Inspection

The bearings were pressed off both journals using a hydraulic bearing puller. The axle identification stampings indicated that the axle was manufactured in August 1992. In addition, the markings indicated that the axle had been normalized and tempered in accordance with AAR grade F axles.



**Fig. 12** One-half of wheelset showing journal fillet (blue arrow), backing ring blend radius (white arrow), cartridge bearing, and end cap (at left)



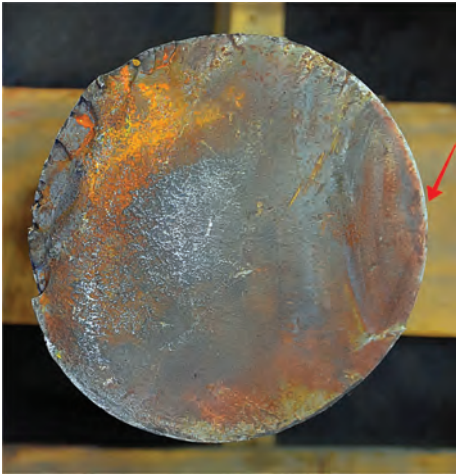
**Fig. 13** Cross-sectional view of fitted backing ring on a fitted axle versus a nonfitted backing ring on a nonfitted axle

The fracture surfaces were cleaned with a nylon brush and soapy water to facilitate a more thorough inspection. Figures 14 and 15 show both halves of the fracture surfaces, with a singular fatigue-crack-initiation site along the outer surface. Beach marks present on the fracture surface confirmed that the crack grew much of the distance through the cross section of the axle prior to sudden rupture and complete separation of the journal. The fracture surface on the wheelset was mechanically damaged postfracture. The postfracture damage destroyed the initiation site of the fracture surface.

The mate journal after bearing removal is shown in Fig. 16 and 17. Pitting is present along the Tectyl application interface. Both journals were cleaned using a solvent to remove the Tectyl from the surface of the journals. Figures 18 and 19 depict the failed



**Fig. 14** Fracture surface of subject journal, axle side (arrow indicates initiation location)



**Fig. 15** Fracture surface of subject journal, bearing side (arrow indicates initiation location)

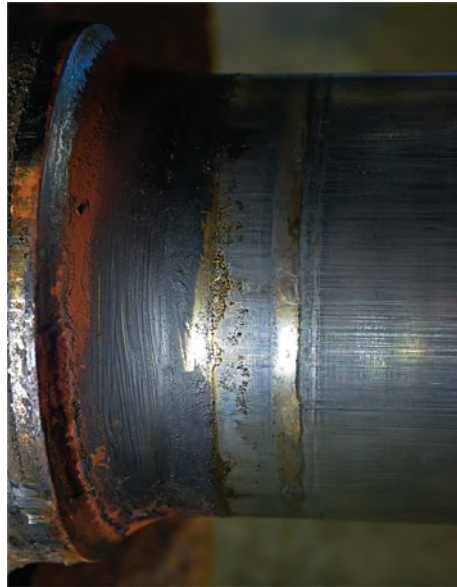


**Fig. 16** Overall view of mate journal after bearing removal

journal outer diameter (OD) after cleaning, with pitting present on the OD near the initiation site. Figure 20 shows the area with the most severe pitting density on the OD of the mate journal fillet.

The journals, along with the fillets, were inspected using a Keyence digital microscope. Figure 21 is a magnified view of the fracture-initiation location. Figures 22 and 23 show secondary cracking emanating from the pitting present on the OD of the fillet journal. The sanding marks bisect the corrosion product within the pit. The mate journal was inspected, and it was noted that pits were present, with the grinding marks also bisecting the corrosion product contained within the pit (Fig. 24). Figures 25 and 26 show pits found on the mate journal fillet where grinding marks follow the edge contours of the bare metal pit walls. Not all of the pits contained debris.

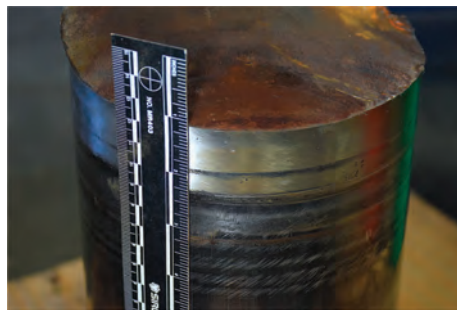
A reconditioned axle will have abrasion marks present along the fillet from the cleaning process. What is unusual is the presence of corrosion pits within the cleaned zone, because all pits are to be removed during axle reconditioning (Ref 6). The presence of abrasion marks within the pit walls (Fig. 25, 26) is a



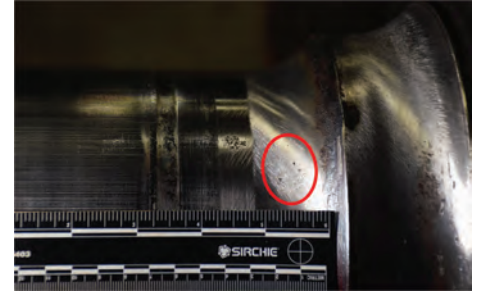
**Fig. 17** Close-up view of mate journal. Left side of image is dust guard, then Tectyl-covered fillet, transitioning to seal wear ring area. Corrosion pits visible on journal diameter onto fillet



**Fig. 18** Failed journal after cleaning at initiation site, with mechanical damage present



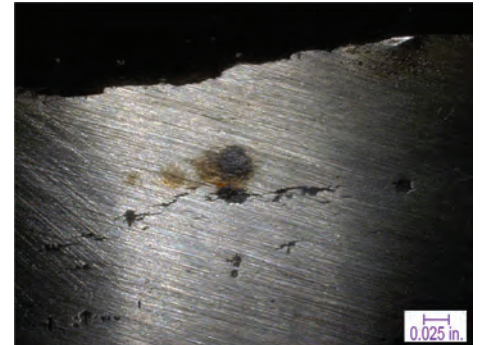
**Fig. 19** Failed journal after cleaning. Pitting is visibly present near the initiation region



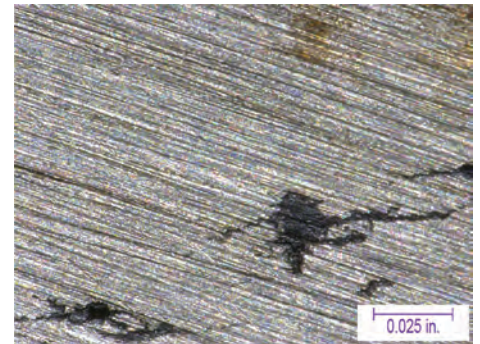
**Fig. 20** Mate journal after cleaning, with pitting present in the fillet region



**Fig. 21** Fracture-initiation location indicated by arrow



**Fig. 22** Secondary cracking emanating from pitting on the failed journal



**Fig. 23** Failed stub secondary cracking. Grinding marks bisect the corrosion product

sign that the pit was present at the time of reconditioning. The effect of pits on the localized stress state of the axle causes localized stresses to exceed the material fatigue (endurance) limit, because progressive fatigue crack propagation morphology was identified on the axle. This is the reason for the AAR MSRP requiring all pitting to be removed.

#### *Formation of Corrosion Pits*

The occurrence of corrosion pitting in axles is not well understood. Common belief is that water ingress into the crevice between the backing ring and journal occurs, which initiates corrosion. Based on this, one would assume that it would be more appropriate to apply copious amounts of Tectyl to the axle to ensure adequate sealing. This too has proven to be problematic, because the Tectyl has been known to migrate along the wheel plate onto the front rim face of the wheel. When the wheelset encounters retarders at a classification yard, the contaminated wheel causes overspeeds to occur due to the reduction in the coefficient of friction on the wheel face.

What does appear to be likely is that holidays form in the Tectyl after application (Fig. 17). The variation in coating thickness may be the source of the pitting corrosion, because exposed steel corrodes, and the steel protected by Tectyl does not corrode.

#### **Discussion**

The question that always arises during a broken journal failure analysis is whether the corrosion pits occurred while the wheelset was in service, or whether, when the axle was reconditioned, preexisting corrosion pits were not fully removed, and consequently, fatigue crack initiation and propagation continued after the wheelset was returned to service. The AAR MSRP requires that all pits be removed; however, there is no specification on how to determine if "all" pits have been removed. Some wheel shops have employed magnification and bright lights to inspect, while others have employed policies that if

corrosion is present, the axle is scrapped. While the latter method is conservative, it does provide a strong position regarding the condition of the axle at time of reconditioning.

The grinding and machining marks present in the debris-filled pits are evidence that some of the corrosion pitting was not fully removed during the axle-reconditioning procedure. The fact that bidirectional machining marks are present suggests that the pitting was sufficiently severe that multiple angles of the belt or disk were attempted to eliminate the existing pits from the fillet. This is evident in Fig. 25 and 26, where the radial marks left by the grinding operation change direction when the pit is encountered.

The wheelset reconditioning operation is as much art and experience as science and specifications. The criticality of axle journal and fillet condition is readily apparent when considering the inability to inspect and considering the likelihood of premature failure if corrosion pits are present on the fillet.

#### **New Assembly Axle Journal Failure after 30 Years of Use**

##### **Introduction**

What made this particular journal failure of interest is that the wheelset was assembled from new components (wheels, axle, bearings)

in December 1990. Therefore, the assembly had been in operation for almost 30 years with no maintenance of the wheelset. More typically, axles and bearings are reconditioned when wheels are reprofiled at the wheel shop.

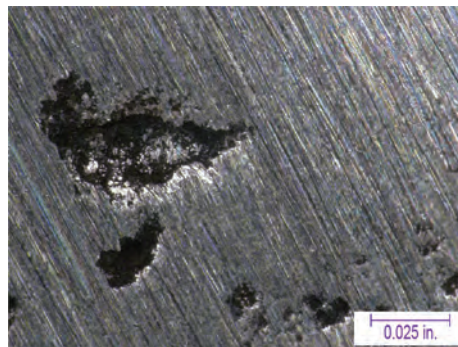
#### **Analysis**

The bearing with the journal section assembled and the mating fracture on the wheelset was photodocumented (Fig. 27–30).

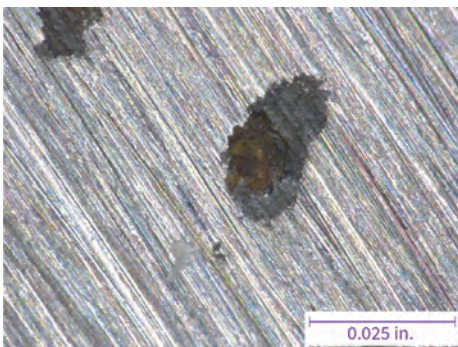
As seen in Fig. 29, the mate bearing was pulled to document the fillet and journal condition. The journal section within the bearing was also pushed out of the bearing. The locking plate on the failed side was documented to record that the bearing was assembled new in December 1990 (Fig. 31). The wheels were cast by Griffin Wheel Company of Bessemer, Alabama, in January 1991; they are grade C, one-wear wheels. The serial numbers were



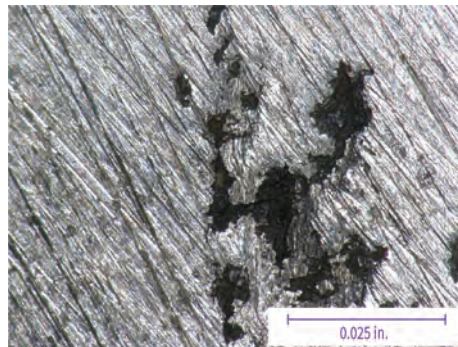
**Fig. 27** Overall view of failed axle journal. The locking plate data are stamped on both bearings to provide information to the bearing installer as to whether the bearing is new or reconditioned, including the month and year of installation. In this case, it was found that the bearing had been installed in new condition in December 1990 by Trinity 54 in Dallas, Texas. These notations are shown in Fig. 31



**Fig. 25** Mate journal pitting with grinding marks present in the bare metal of the pit



**Fig. 24** Mate axle with pitting corrosion product bisected by grinding marks



**Fig. 26** Mate journal pit with grinding marks following the contours of the pit walls



**Fig. 28** Fracture surface on axle side. Note that the backing ring remains in place on the axle due to the fitted backing ring

04640 for the mate-side wheel and 03456 for the failed-side wheel.

The locking plate data are stamped on both bearings to provide information to the bearing



**Fig. 29** Mating side of axle with bearing removed. The failed journal was still within the bearing assembly, as seen in Fig. 30. A distinct population of ratchet marks is observable around the circumference of the fracture surface. The ratchet marks indicate the combination of crack planes



**Fig. 30** Fractured axle journal within bearing. The end cap was removed from the failed side. A noticeable amount of red rust was present on the interior of the end cap, as seen in Fig. 32. The journal stub was pressed out of the bearing. Present on the end of the journal was the axle fabrication data, shown in Fig. 33. The axle was manufactured by Standard Forged Products, McKees Rocks Forgings of McKees Rocks, Pennsylvania, in June 1990



**Fig. 31** Cap screw locking plate. Blue arrow indicates "N" for new, while white arrows indicate month and year (December 1990)

installer as to whether the bearing is new or reconditioned, including the month and year of installation. In this case, it was found that the bearing had been installed in new condition in December 1990 by Trinity 54 in Dallas, Texas. These notations are shown in Fig. 31.

The failed journal was still within the bearing assembly (Fig. 30). Nearest to the fracture surface are the inner seal wear ring, the seal housing, and the chipped bearing cup. The seal wear ring would have been pressed into the back side of the backing ring (Fig. 28). A distinct population of ratchet marks is observable around the circumference of the fracture surfaces. The ratchet marks indicate the combination of cracking planes. The fracture surfaces showed fracture initiation from multiple locations around the circumference of the journal.

The end cap was removed from the failed side. A noticeable amount of red rust was present on the interior of the end cap (Fig. 32). The journal stub was pressed out of the bearing. Present on the end of the journal was the axle fabrication data (Fig. 33). The axle was manufactured by Standard Forged Products, McKees Rocks Forgings of McKees Rocks, Pennsylvania, in June 1990.

The bearing components were degreased and cleaned in solvent. Beyond typical fretting



**Fig. 32** Inside of subject bearing end cap



**Fig. 33** Axle journal end stampings. Clockwise: manufacturer's marks (MRF, McKees Rocks Forgings; date of manufacture, June 1990), heat number, and serial number. The bearing components were degreased and cleaned in solvent. Beyond typical fretting wear seen in the ridges worn into the inner cone (race), as seen in Fig. 34, no negative conditions were noted within the components of either bearing

wear seen in the ridges worn into the inner cone (race) (Fig. 34), no negative conditions were noted within the components of either bearing.

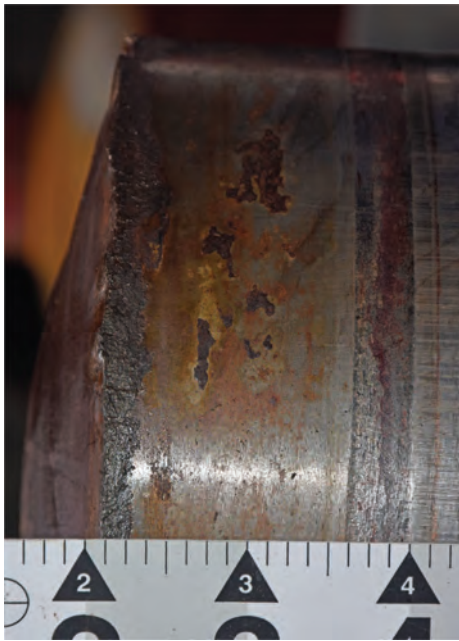
The fractured journal was also cleaned in mineral spirits and then cut approximately 25 mm (1 in.) from the fracture surface. This was performed to enable the area adjacent to the fracture to be inspected. Prior to inspection using a digital light microscope, the section was soaked in Evapo-Rust (Harris International) to remove the tightly adhering red rust adjacent to the fracture surface (Fig. 35). An example of the surface condition of the axle is shown in Fig. 36. The fracture is to the viewer's left, and the concentric lathe tool cutting marks can be seen as vertical parallel grooves farther to the right of the pitting.

Increased magnification inspection of the surface revealed sharp-edged pits with well-defined edges and, to a greater extent, circular pits on the axle (Fig. 37). Note that the lathe tool markings are intersected by the corrosion pits. This is logical because the axle was first machined and then, during operation, corrosion occurred that intersected the groove marks. This observation is obvious, knowing a priori that the axle had never been reconditioned.

During analysis of journals that have failed with reconditioned bearings and axles, situations arise where the edges of the pits are dulled or abraded by incomplete pit removal during the reconditioning process. Because this axle was a new application, it provides a visual basis for the appearance of corrosion pitting of a journal that has never undergone a reconditioning process. In addition to the lack of abrasion, there are also locations visible where rust debris remained embedded in



**Fig. 34** Seal wear ring (inverted) below seal housing. Fretting wear can be seen on the inner cone at the bottom of the stack. The fractured journal was also cleaned in mineral spirits and then cut approximately 25 mm (1 in.) from the fracture surface. This was performed to enable the area adjacent to the fracture to be inspected. Prior to inspection with a digital light microscope, the section was soaked in Evapo-Rust to remove the tightly adhering red rust adjacent to the fracture surface, as seen in Fig. 35. An example of the surface condition of the axle is shown in Fig. 36. The fracture is to the viewer's left, and the concentric lathe tool cutting marks can be seen as vertical parallel grooves farther to the right of the pitting



**Fig. 35** Side view of axle stub, with fracture on left, showing heavy buildup and groove from seal wear ring

the pits (Fig. 38, 39). It is noteworthy that there are no abrasion marks present in the embedded debris. In reconditioned bearings, it has been observed that the embedded debris contains abrasion markings due to the abrasion that occurs during the reconditioning process. Of course, if properly reconditioned, there should be no pitting present in the fillet area of the journal.

Greater magnification inspection around the circumference of the axle revealed additional fatigue cracks (secondary) near the fracture surface (Fig. 39).

### Results and Discussion

Advanced pitting corrosion occurred between the fitted backing ring and the axle journal. The pitting acted as stress-concentration locations when the axle rotated under load. Fatigue cracks initiated in multiple locations around the circumference of the journal in the area of corrosion, leaving the distinctive ratchet marks observed on both fracture surfaces. The fatigue cracks then coalesced into a primary crack, which progressively propagated through the cross section of the axle and finally failing in bending overload, causing the derailment. In contrast, journals that have

failed after reconditioning often do not exhibit multiple ratchet marks; rather, the fracture initiates from a singular location or area of the journal with minimal ratchet marks present.

Notably, the mate axle journal had only minimal rust behind the backing ring, and the rust inhibitor applied to the axle journal fillet prior to bearing application appeared to be in good condition. No conclusive evidence was identified as to why there was a notable amount of corrosion present under one backing ring versus the other.

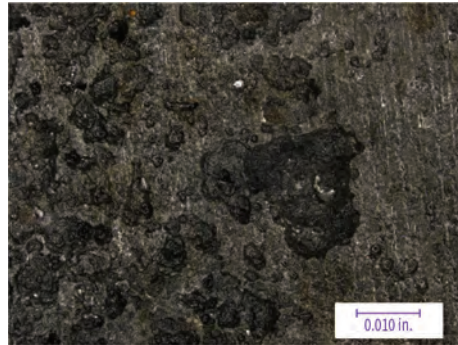
### Conclusions

The axle journal failed due to the presence of corrosion pits around the circumference of the journal below the backing ring and seal wear ring. The corrosion pits raised the localized bending stress above the endurance limit and initiated multiple fatigue cracks on the circumference of the journal. The fracture surfaces were consistent with the morphology of rotating-bending fatigue.

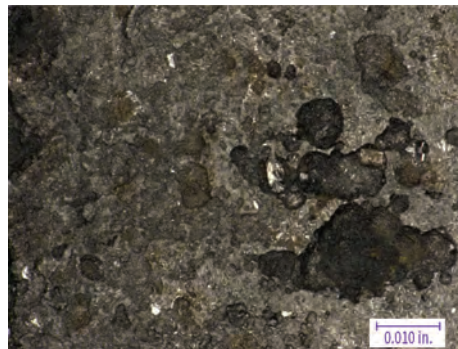
The bearing was applied new to a new axle in December 1990; the installation was in operation for more than 30 years with no maintenance. The mate side of the axle still had intact rust inhibitor on the journal and a minimal amount of corrosion present on the end cap, whereas the subject side contained a notable amount of corrosion on the end cap and on the journal near the pitting and fracture area. No evidence was discovered to explain why one journal had corrosion on the journal fillet while the other journal was in a noncorroded condition.

### Bearing Failures

When railroad components fail, the component often becomes substantially more damaged before the failure is detected. This especially holds true for overheated freight car bearings, because they generate sufficient heat to damage all components involved, including the axle journal, the location where the bearing is applied, causing what is known



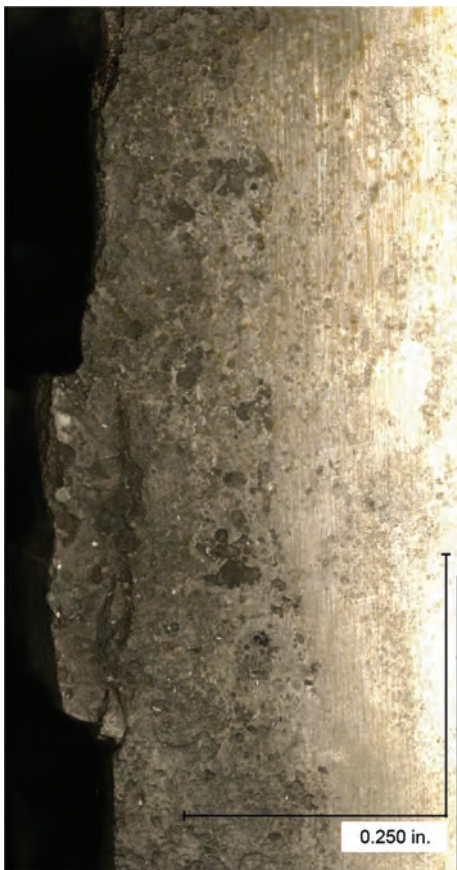
**Fig. 37** General pitting present in the axle. Note parallel lathe tooling marks interrupted by corrosion pits



**Fig. 38** Pronounced pitting present on journal surface. Note the corrosion material embedded in some of the pits



**Fig. 39** Secondary fatigue cracks present near primary fracture surface



**Fig. 36** Overall view of pitting near fracture surface

as burnoff. A teardown of the mating bearing to inspect for damage is often performed to help infer the cause of failure of the burned-off bearing.

The general arrangement of the axle journal and bearing is shown in Fig. 40. Note that the double-taper roller bearing inner races as well as the seal wear rings are interference-fit to the axle journal. After pressing the bearing onto the axle journal, an end cap and three cap screws maintain the compressive load onto the bearing stack.

The aftermath of failed bearings can indicate the speed at which the failure occurred and the resulting location where the heat was applied. The majority of roller bearing failures occur on the most highly stressed part of the bearing, which is the inner cone, because the stresses are greatest closest to the wheel. When the inner cone fails, the heat generated causes rapid heating and failure of the journal. Because the inner race is the high-stress-concentration location, the axle fails quicker, and the remaining journal stub is typically shorter. When the outboard cone fails first, the entire journal is heated until the stresses at the point of highest stress (next to the wheel) cause the journal to fail. Because heating occurs from the outside in, the journal is eventually deformed and elongated, resulting in a long, tapered journal stub. Figures 41 and 42 show the differences in location when initial failure occurs at the inner race location versus the outer race location.

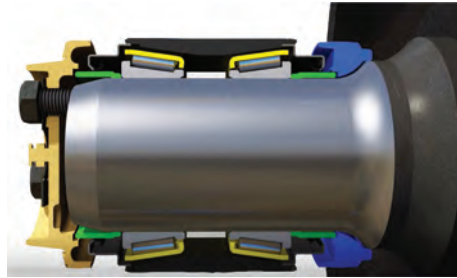
There are a few common causes of bearing failures in freight car service. Because railroad bearings are meant to last for many years with no maintenance, the more common failures that are not related to wear over time, such as lubrication breakdown, general spalling of components, or uneven loading of the bearing, are not covered. Three common types of bearing failures seen in rail service are loose components, water ingress, and failures due to high impacts, resulting in broken internal components.

### Loose Components

When disassembling a bearing, it is important to examine the front hub and plate of the wheel, because it often can contain grease that escaped from the bearing, a sign that the grease seal was compromised. An indication of bearing component stack loss is that the cap screw torques holding the bearing end cap in place are low when compared to the installation torque. The normal bolt torques are in the range of 290 to 420 lb/ft, depending on the bearing size. If the torques are less than the specified values, the preload of the bearing cartridge assembly will result in poor lateral retention and end loading of rollers, which can lead to overheating. Tapered roller bearings are designed to operate on the crown, or center, or the roller. When the clamp load is lost and the bearing

rollers are end loaded, the ends of the bearings will spall and fail (Fig. 43).

While the cap screws were likely installed with the proper torque, over time and usage the flexing of the axle journal will lead to fretting damage between the bearing cones and seal wear rings. When the components of the bearing begin to lose the lateral clamp load, movement of the components can exacerbate the issue by wearing further. Back face wear of the roller cones occurs when the cone interacts with the seal wear ring, wearing a step into the cone. The loss of material reduces the lateral preload. Figure 44 shows the step created from back face wear in the roller cone.



**Fig. 40** Sectional view of the cartridge bearing and related components on the axle journal



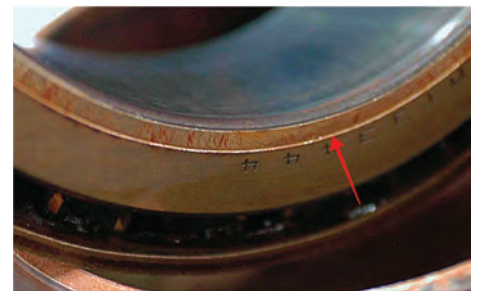
**Fig. 41** Burned-off journal stub. The short, rough stub indicates high heat over a short time period and is associated with an inner cone failure



**Fig. 42** Burned-off journal stub. The long, thinned-out journal indicates that the bearing was hot over a long time period and is associated with an outer cone failure



**Fig. 43** Tapered roller bearing fracture from end loading due to loss of lateral clamp force



**Fig. 44** Back face wear of a bearing cone. Arrow indicates the step present from wear against the seal wear ring

to the failed side had a cone that showed fretting wear (Fig. 45) and discolored rollers. The failed-side bearing had fused together the inner bearing components (Fig. 46).

### Water Ingress

Another common failure mode is water ingress into freight rail bearings, causing accelerated failure. The most prevalent sign of water ingress is free water during the bearing removal and disassembly process, followed by emulsified grease. Grease that has had water emulsified into it appears milky (Fig. 47) rather than the normal black or yellow. It should be noted that the aforementioned fretting damage stains the grease a red-rust color, but this coloring could also be due to rusting of components. Therefore, caution should be practiced to evaluate the seal condition for wear before concluding that the grease discoloration is due to water ingress. Water intrusion into the bearing results in lubrication breakdown. When teardowns of bearings are performed, the amount of grease present is measured by weight; if the amount of grease is over the specified amount, it can be assumed that at least that much water has been trapped. Water ingress often happens from worn-out seals, fit, or the cars were exposed to flood water.

Water entrainment in the grease leads to water etching of the rolling bearing components (Fig. 48). Water etching leads to spalling

of the races and rollers. These spalled fragments that are now also entrained in the grease act as an abrasive, wearing rollers and cages. If the cage roller pockets wear sufficiently, the rollers skew in the pockets and slide rather than roll. Water etching can result in cracking of the internal components of the bearings if the condition is pronounced or if sufficient time has transpired to reach a critical point. Figure 49 shows microcracking present with water etching and rust on the cup of a bearing.

### Wheel Impacts

Often, the mate bearing does not provide reasons for the failed bearing. If end cap retaining bolt torques and bearing surface conditions do not cause concern, one could look to the wheel on the side of the failed bearing. Tread shelling or other wheel defects can cause elevated impact loading to the bearing due to an out-of-round wheel condition. Railroads employ wheel impact load detectors installed at various points throughout their networks to help identify wheelsets that need maintenance or replacement.

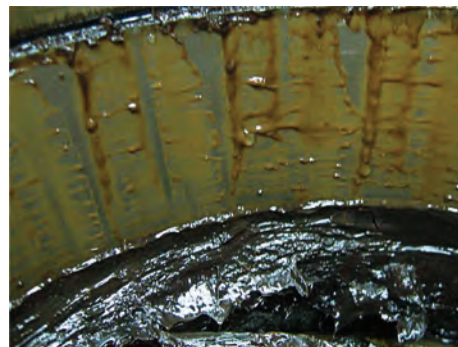
The impact loads can cause the bearing cage to fail in fatigue, resulting in skewed and seized rollers and an eventual failure. The cage pocket corners are the locations where the cracking occurs. Other signs of cage failure



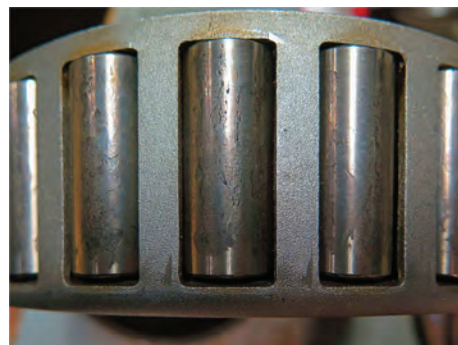
**Fig. 45** Fretting wear on the inside diameter of a bearing with a spun cone



**Fig. 46** Damage from a spun cone bearing failure. All internal components were fused together



**Fig. 47** Grease condition during bearing disassembly with water ingress. Free water was noted, and milky grease is present along the rolling-contact surfaces



**Fig. 48** Water etching present on rollers

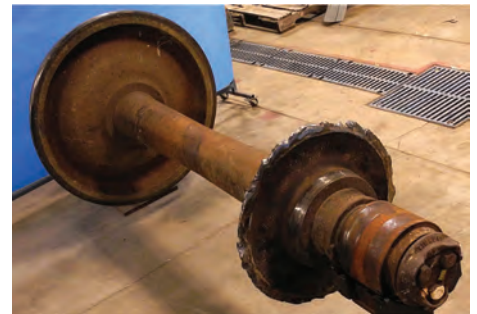
are wear marks present on the cage, due to misalignment of the rollers.

### Wheel Failure due to Corrosion

A failed wheel was found when a dragging equipment warning occurred. The train was stopped, and it was noted that a wheel was broken but riding on the rail on the hub. The failed wheel was a J33 design, forged from grade C steel and manufactured in September 1992. The locking plates on the bearings indicated that the bearings were rebuilt and installed in March 1994. Figures 50 and 51



**Fig. 49** Water etching with microcracking present on the cup of a bearing



**Fig. 50** As-received wheelset with portion missing



**Fig. 51** As-received rim sections viewed from the back rim face. Circled area indicates initiation region

show the wheelset and rim sections in the as-received condition.

### Analysis

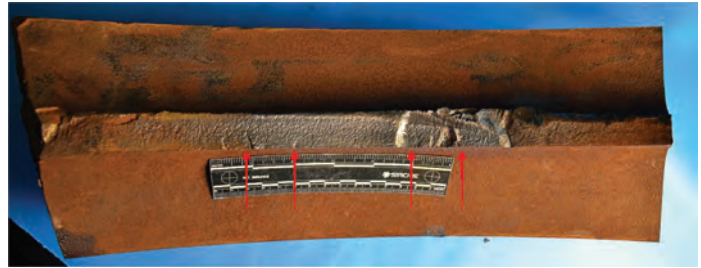
The wheelset was inspected, and it was noted that the hub portion had ridden the rail for some time. A view of the hub is shown in Fig. 52, with wear to the hub indicated by an arrow. The plate portion extending from the hub had been damaged from impacts to the track structure while rolling on the rail. The fracture surfaces of the hub side were indicative of mechanical damage and thus did not provide any relevant information.

The rim portions are shown in Fig. 51, with a fracture-initiation area indicated by a red circle. This portion was sectioned and cleaned using an ultrasonic cleaner and a mild detergent. Figure 53 shows an overall view of the fracture surface after cleaning, with the initiation location indicated by arrows. Figure 54 is a closeup of the area at the left in Fig. 53. The initiation locations were characterized by small, flat half-circles indicative of fatigue fracture before transitioning into rapid failure. The presence of ratchet marks, the steps along the fracture surface, indicates that several initiation sites combined into one fracture surface. The front plate face was inspected for the possibility of a plate strike; no signs of indentations were noted in the old break region of the rim sections. The plate surface near the fracture-initiation locations is shown in Fig. 55; some corrosion-induced pitting was found to be present.

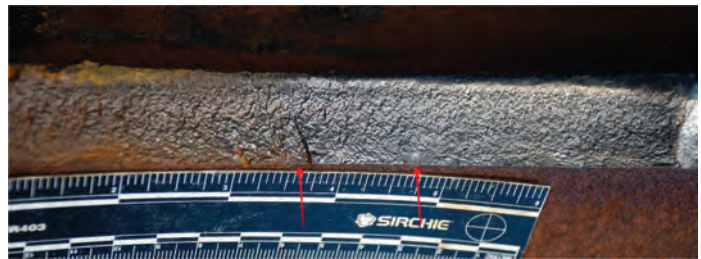
A section was removed from the area shown in Fig. 54 and inspected using a digital microscope. The fracture surface had been rubbing for some time, causing loss of definition of fine features. Figure 56 shows a closeup of one of the crack-initiation locations. Sections were removed from the area shown in Fig. 56 and mounted and polished according to standard metallographic procedures. The samples were examined in the as-polished state. It was observed that the cracks had started at pits near the fracture-initiation location (Fig. 57–59). The samples were etched with a 2% nital solution to reveal the microstructure, which was consistent with pearlite, as expected. It was

noted that the surfaces had some decarburization present. The amount of decarburization present on the fracture-initiation side was less than that on the opposite side of the wheel (Fig. 60, 61).

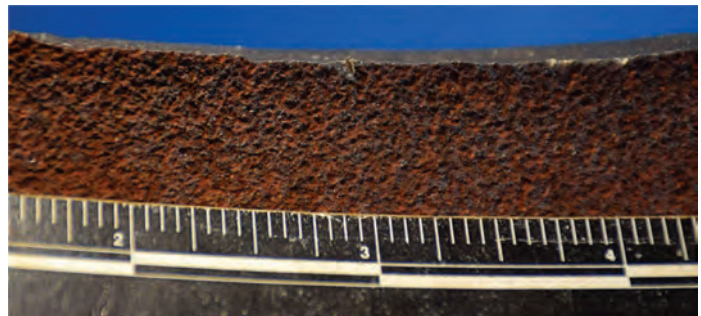
Chemical composition analysis was performed using an ARL optical emission spectrometer. Table 1 contains the results of the chemical analysis, along with the specifications



**Fig. 53** Overall view of fracture surface sectioned from rim shown in Fig. 51. Arrows indicate fracture-initiation locations



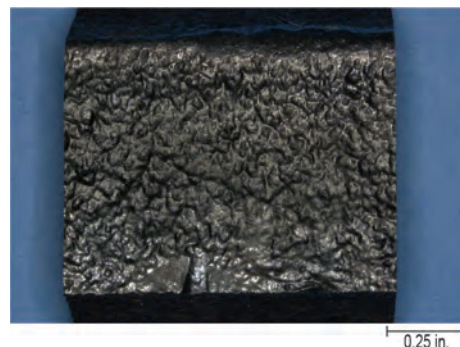
**Fig. 54** Closeup of the two fracture-initiation locations at the left in Fig. 53



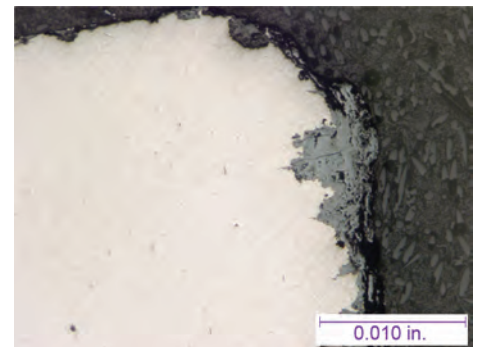
**Fig. 55** Plate surface condition around the fracture-initiation locations. No evidence of a plate strike was present



**Fig. 52** Hub and partial plate section of failed wheel. Arrow shows the wear from riding on the rail



**Fig. 56** Closeup of fracture-initiation location



**Fig. 57** Secondary cracking present within pits. Fracture surface is toward the top of the image

from AAR M-107/M-208 for class C wheels. The chemical composition of the wheel was within AAR specification.

Hardness testing was performed on a section of the wheel. Table 2 contains the results of the testing along with the specifications from AAR M-107/M-208 for class C wheels. The hardness results were within AAR specification.

Discussion and Conclusion

The curved plate wheel design used by the AAR incorporates the ability of the wheel

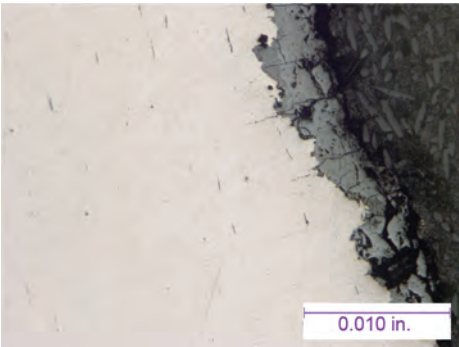


Fig. 58 More secondary cracking from an area with surface corrosion

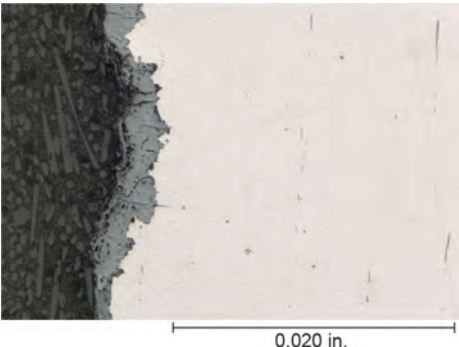


Fig. 59 Secondary cracking present within corrosion pits

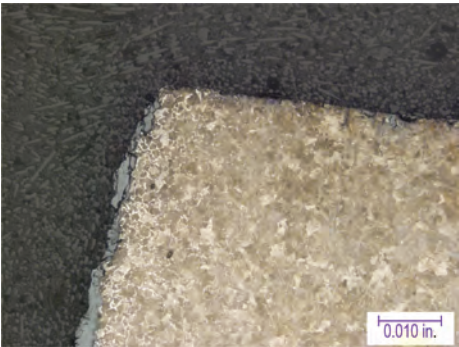


Fig. 60 Fracture initiation is toward the top of the image. Some decarburization is present along the forging surface

plate to flex. The flexing causes one surface of the wheel plate to experience tension when the opposite side experiences compression. This is within expected performance; however, wheel plates that contain pronounced corrosion pitting have been found to experience plate-initiated fatigue cracking. Thus, the wheel failed due to a fatigue crack that initiated from a corrosion pit on the back plate face of the

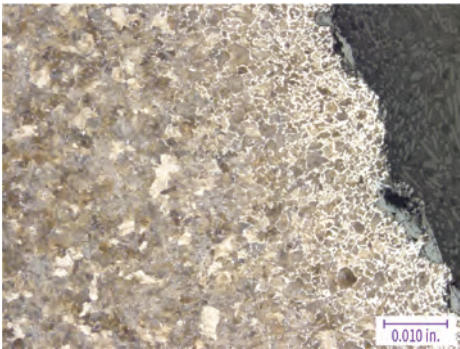


Fig. 61 Opposite side of plate to initiation. More decarburization is present along the forging surface

Table 1 Chemical analysis in accordance with ASTM International Standard E415

|                                        | Chemical composition, wt% |      |       |       |      |      |      |      |       |      |       |      |       |
|----------------------------------------|---------------------------|------|-------|-------|------|------|------|------|-------|------|-------|------|-------|
|                                        | C                         | Mn   | P     | S     | Si   | Ni   | Cr   | Mo   | V     | Cu   | Al    | Ti   | Nb    |
| AAR M-107/M-208 class C                |                           |      |       |       |      |      |      |      |       |      |       |      |       |
| Maximum                                | 0.77                      | 0.90 | 0.030 | 0.040 | 1.00 | 0.25 | 0.25 | 0.10 | 0.040 | 0.35 | 0.060 | 0.03 | 0.05  |
| Minimum                                | 0.67                      | 0.60 | ...   | 0.005 | 0.15 | ...  | ...  | ...  | ...   | ...  | ...   | ...  | ...   |
| Wheel section                          | 0.75                      | 0.73 | 0.015 | 0.019 | 0.26 | 0.06 | 0.10 | 0.02 | 0.007 | 0.13 | 0.003 | 0.02 | 0.001 |
| AAR, Association of American Railroads |                           |      |       |       |      |      |      |      |       |      |       |      |       |

AAR, Association of American Railroads

Table 2 Hardness testing results

|                         | Test readings, HBW | Notes                 |
|-------------------------|--------------------|-----------------------|
| AAR M-107/M-208 class C | ...                | 321–363 HBW specified |
| Wheel section           | 341, 302, 341      | 328 HBW average       |

AAR, Association of American Railroads



Fig. 62 As-received wheelset with damage present to the running surface

wheel. Secondary cracks from corrosion pits were also noted in the area. Specific rules within the AAR were developed to ensure that wheels are not stored horizontally when exposed to the elements. This is done to prevent pooling of water and the formation of corrosion product.

Wheel Failure due to Thermal Cracks and Subsurface Fatigue

Figure 62 shows the as-received failed wheelset. The failed wheel, serial number 23662, was a one-wear curved plate, CH36 design with a heat treatment of “C,” cast in November 2008. The mate wheel, serial number 23767, was of the same design and manufacture as the failed wheel. The wheels were mounted in April 2009.

Analysis

Visual inspection of the wheel showed cracking within the tread in four areas (Fig. 63). The cracking was similar to a shattered rim; however, the fatigue cracking progressed in a stepped manner. Close-up views of the fractures from the front rim face and the running surface are shown in Fig. 64 to 70.

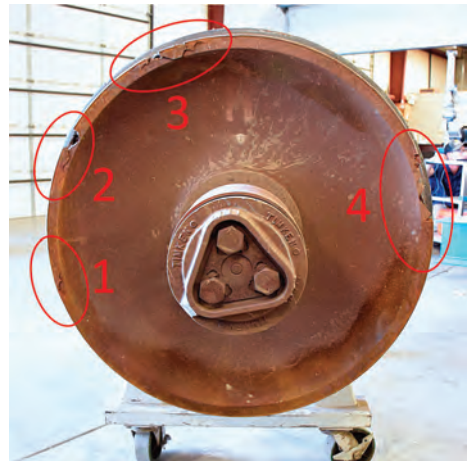


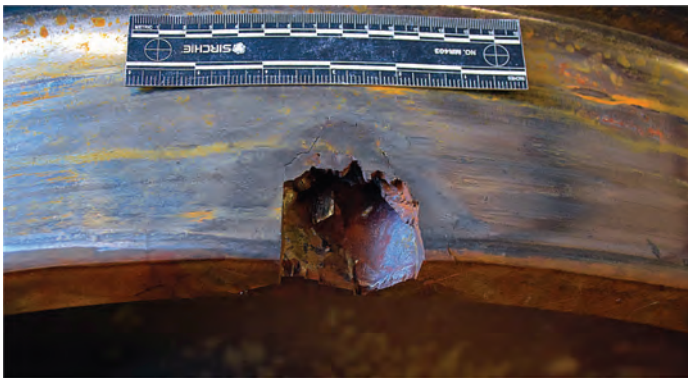
Fig. 63 End view of the wheelset. Fracture locations are numbered and indicated by ovals



**Fig. 64** Close-up end view of areas 1 and 2 from Fig. 63



**Fig. 65** Top view of damage to area 1 from Fig. 63



**Fig. 66** Top isometric view of damage to area 2 from Fig. 63



**Fig. 67** Close-up end view of area 3 from Fig. 63



**Fig. 68** Top view of damage to area 3 from Fig. 63



**Fig. 69** Close-up end view of area 4 from Fig. 63

Some of the cracks were quite visible to the naked eye; however, there were subtle indications that warranted ultraviolet magnetic-particle inspection to be performed on the wheel. Widespread cracking was identified over the circumference of the wheel tread. The cracks were similar to thermal cracks but extended farther into the running surface and, frequently, at an angle up to  $20^\circ$  with respect to the tread surface. The cracking is indicated by arrows in Fig. 71 to 73.

After examining the fracture and crack locations, a section of the rim between fracture areas 2 and 4 was removed from the wheel. The wheel section was further thinned to allow the cracks to be opened. Figure 74 shows the results of sectioning: opened fracture surfaces in addition to a hot macroetched section of wheel tread approximately 9.5 mm ( $3/8$  in.) from the front rim face. The macroetch was performed when difficulties arose in opening the running surface cracks; the macroetch sample is shown in Fig. 75. A close-up view of the branched cracking present in the macroetch is shown in Fig. 76.



**Fig. 70** Top view of damage to area 4 from Fig. 63

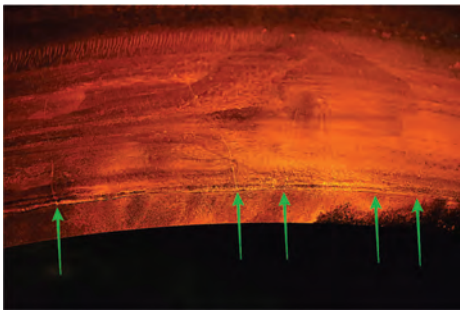


Fig. 71 Magnetic-particle testing results from near fracture area 1. Arrows indicate crack locations

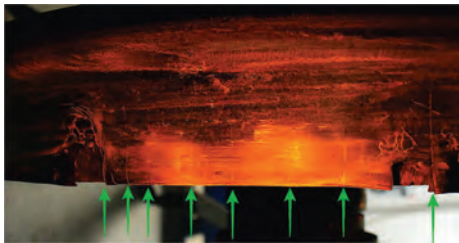


Fig. 72 Magnetic-particle testing results from between fracture areas 2 and 3. Arrows indicate crack locations

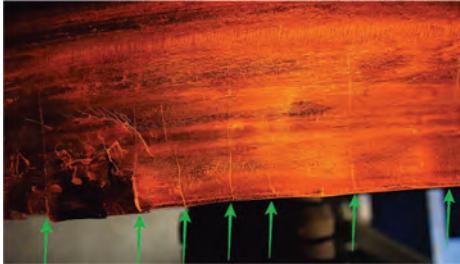


Fig. 73 Magnetic-particle testing results from near fracture area 3. Arrows indicate crack locations



Fig. 74 Analyzed rim section showing fracture locations and macroetch performed on a section

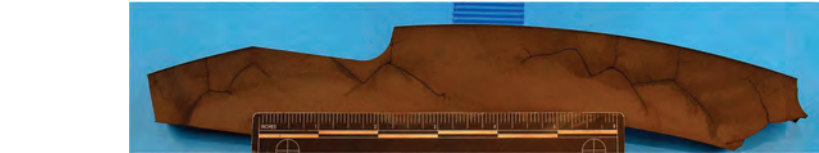


Fig. 75 Macroetch results of section taken 9.5 mm (3/8 in.) from the rim face

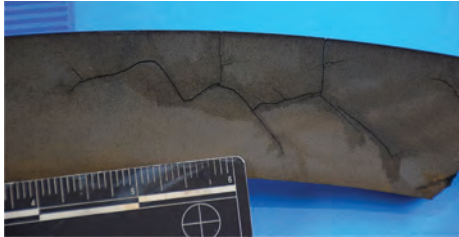


Fig. 76 Closeup of branched cracking present in the macroetch sample



Fig. 77 Circle indicates initiation location on fracture surfaces



Fig. 78 Opposite fracture surface. Arrow indicates initiation location

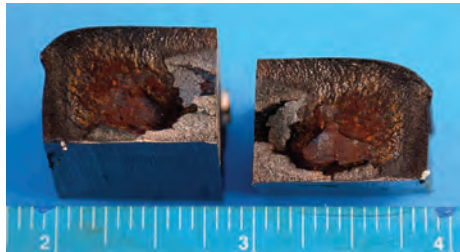


Fig. 79 Thermal crack revealed from further sectioning of the wheel tread sections

Further sectioning of the rim sample was performed, and additional fracture surfaces were revealed. An extensive fatigue crack network, with typical beach marks originating from peaks or one of the triple points, is shown in the macroetch in Fig. 76. One of the fracture surfaces had been open to the atmosphere, and some surface rust was present (Fig. 77). The opposing fracture surface was still bare metal, indicating that the fractures had not been exposed to the atmosphere (Fig. 78).

The fracture surfaces revealed in Fig. 77 and 78 were subsurface cracks and were not the cracks found on the running surface from the magnetic-particle inspection. Further sectioning exposed a section of one of the cracks found on the surface (Fig. 79). The dark surfaces are indicative of a thermal crack.

An ARL optical emission spectrometer was used to perform chemical analysis on a section of the wheelset. The chemical composition of the wheel was compared to that of AAR M-107/M-208 for class C wheel steel. The material was confirmed to be within specification, as shown in Table 3.

A cross section of the wheel was ground smooth for Rockwell hardness mapping in accordance with AAR M-107/M-208, Fig. 11.1. The specified hardness range was 30 to 42 HRC. Table 4 contains the results of the hardness

Table 3 Chemical analysis in accordance with ASTM International Standard E415

|                                        | Chemical composition, wt% |      |       |       |      |      |      |      |       |      |       |      |      |
|----------------------------------------|---------------------------|------|-------|-------|------|------|------|------|-------|------|-------|------|------|
|                                        | C                         | Mn   | P     | S     | Si   | Ni   | Cr   | Mo   | V     | Cu   | Al    | Ti   | Nb   |
| AAR M-107/M-208 class C wheel          |                           |      |       |       |      |      |      |      |       |      |       |      |      |
| Maximum                                | 0.77                      | 0.90 | 0.030 | 0.040 | 1.00 | 0.25 | 0.25 | 0.10 | 0.040 | 0.35 | 0.060 | 0.03 | 0.05 |
| Minimum                                | 0.67                      | 0.60 | ...   | 0.005 | 0.15 | ...  | ...  | ...  | ...   | ...  | ...   | ...  | ...  |
| Wheel section                          | 0.75                      | 0.81 | 0.018 | 0.018 | 0.50 | 0.06 | 0.10 | 0.02 | 0.003 | 0.14 | 0.008 | 0.00 | 0.00 |
| AAR, Association of American Railroads |                           |      |       |       |      |      |      |      |       |      |       |      |      |

AAR, Association of American Railroads

mapping of the wheel. The wheel met the AAR hardness specifications.

Two sections were removed from the left side of the macroetch sample (Fig. 75). These samples were mounted, polished, and etched with 2% nital in accordance with standard metallographic procedures. The samples were examined, and it was noted that the texture of the steel was more columnar toward the running surface and became more equiaxed where the branched cracks were farther from the running surface. This observation is best seen in Fig. 80 and 81; both images are overall photomerges of the metallographic samples. The texture is likely the result of solidification of the steel during casting of the wheel. The microstructure observed at the surface of both samples consisted of tempered martensite (Fig. 82). Martensite is not permitted in freight car wheels; the core microstructure was consistent with pearlite (Fig. 83).

Microhardness profiles were performed on each metallographic sample, using a Leco M-

400-G2 hardness tester with 500 gf and a diamond pyramid indenter for Knoop hardness (HK) readings. The results of the microhardness profiles, along with the Rockwell C scale conversions, are shown in Table 5. Surface hardness in excess of 50 HRC was noted.

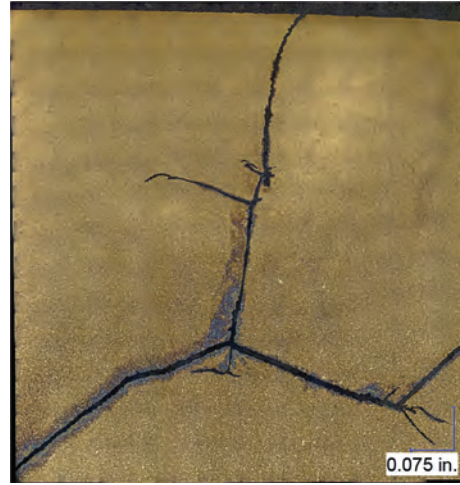
A section the fracture surface containing the vertical crack that was shown on the running

surface was cleaned and examined using scanning electron microscopy. The fatigue cracking on the sample initiated from an area of intergranular cracking, indicative of thermal cracking. The intergranular cracking can be seen in Fig. 84.

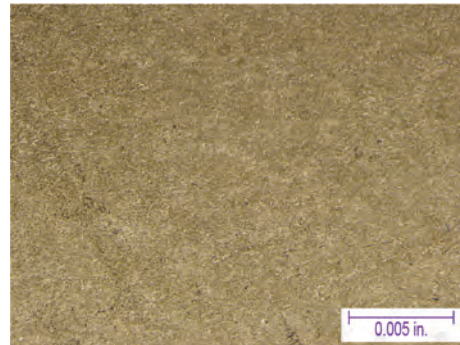
**Table 4 Hardness map in accordance with AAR M-107/M-208**

| Hardness, Rockwell C |     |               |                                                                               |                                                       |
|----------------------|-----|---------------|-------------------------------------------------------------------------------|-------------------------------------------------------|
| Depth                |     | Flange throat | Midtread (77.5 mm, or 3 <sup>1</sup> / <sub>16</sub> in., from back rim face) | Tread edge (12.5 mm, or 1/2 in., from front rim face) |
| mm                   | in. |               |                                                                               |                                                       |
| 3.2                  | 1/8 | 37.2          | 37.5                                                                          | 41.2                                                  |
| 6.5                  | 1/4 | 37.2          | 37.6                                                                          | 39.8                                                  |
| 9.5                  | 3/8 | 36.3          | 34.2                                                                          | 36.1                                                  |
| 12.7                 | 1/2 | 37            | 34.8                                                                          | 35.5                                                  |
| 15.9                 | 5/8 | 35.6          | 35.4                                                                          | 36.7                                                  |
| 19.1                 | 3/4 | 35.2          | 35.6                                                                          | 36.5                                                  |
| 22.2                 | 7/8 | 33.6          | 35.6                                                                          | 35.6                                                  |
| 25.4                 | 1   | 35.7          | 35.9                                                                          | 35.8                                                  |

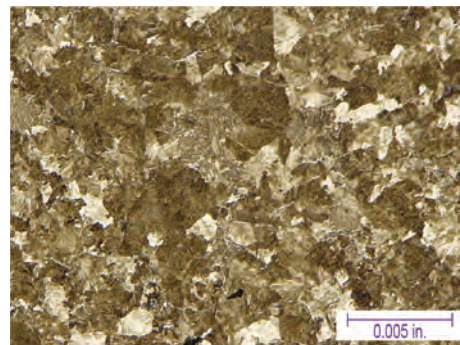
AAR, Association of American Railroads



**Fig. 81** Overall photomerge view of metallography sample 2



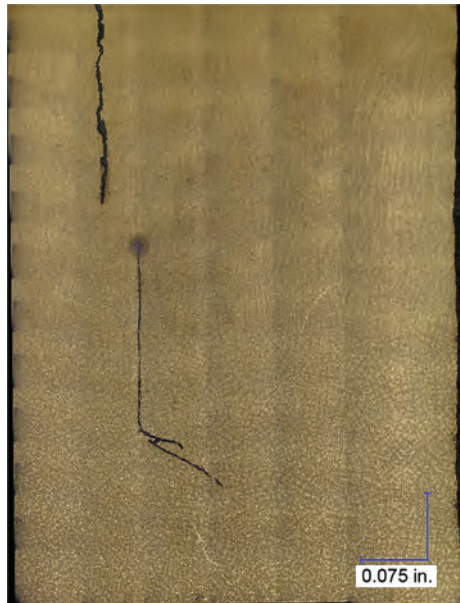
**Fig. 82** Representative surface microstructure consisting of tempered martensite



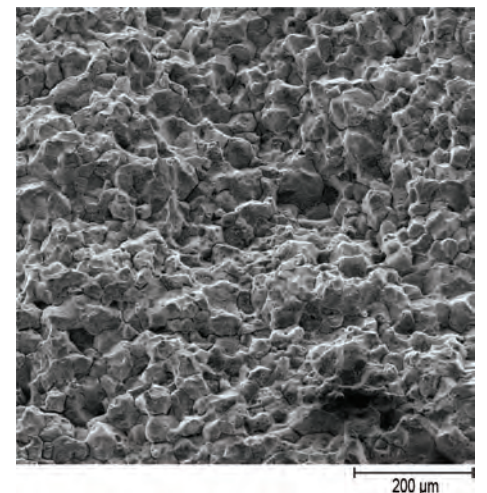
**Fig. 83** Representative core microstructure consisting of pearlite

**Table 5 Microhardness profiles**

| Spacing |      | Mount 1 |     | Mount 2 |     |
|---------|------|---------|-----|---------|-----|
| mm      | in.  | HK      | HRC | HK      | HRC |
| 0.51    | 0.02 | 558     | 51  | 557     | 51  |
| 1.02    | 0.04 | 523     | 49  | 519     | 49  |
| 1.52    | 0.06 | 497     | 47  | 482     | 46  |
| 2.03    | 0.08 | 452     | 44  | 463     | 45  |
| 2.54    | 0.10 | 455     | 44  | 434     | 43  |
| 3.05    | 0.12 | 430     | 42  | 424     | 42  |
| 3.55    | 0.14 | 413     | 41  | 408     | 40  |
| 4.06    | 0.16 | 445     | 44  | 417     | 41  |
| 4.57    | 0.18 | 443     | 43  | 408     | 40  |
| 5.08    | 0.20 | 432     | 43  | 414     | 41  |
| 5.58    | 0.22 | 443     | 43  | 442     | 43  |
| 6.09    | 0.24 | 440     | 43  | 426     | 42  |
| 6.60    | 0.26 | 454     | 44  | 432     | 43  |
| 7.11    | 0.28 | 414     | 41  | 438     | 43  |
| 7.61    | 0.30 | 430     | 42  | 452     | 44  |
| 7.87    | 0.31 | 428     | 42  | 444     | 43  |
| 8.38    | 0.33 | 433     | 43  | 440     | 43  |
| 8.88    | 0.35 | 415     | 41  | 415     | 41  |
| 9.39    | 0.37 | 392     | 39  | 415     | 41  |
| 9.90    | 0.39 | 426     | 42  | 411     | 41  |
| 10.41   | 0.41 | 424     | 42  | 410     | 41  |
| 10.91   | 0.43 | 425     | 42  | 418     | 41  |
| 11.42   | 0.45 | 394     | 39  | 415     | 41  |
| 11.93   | 0.47 | 424     | 42  | 394     | 39  |
| 12.44   | 0.49 | 423     | 42  | 388     | 39  |
| 12.94   | 0.51 | 406     | 40  | 393     | 39  |
| 13.45   | 0.53 | 410     | 41  | 394     | 39  |
| 13.96   | 0.55 | 408     | 40  | 386     | 38  |
| 14.47   | 0.57 | 400     | 40  | 387     | 39  |
| 14.98   | 0.59 | 415     | 41  | 402     | 40  |
| 15.48   | 0.61 | 399     | 40  | 447     | 44  |
| 15.99   | 0.63 | 411     | 41  | 393     | 39  |
| 16.50   | 0.65 | 384     | 38  | 398     | 40  |
| 17.01   | 0.67 | 402     | 40  | 383     | 38  |
| 17.51   | 0.69 | 426     | 42  | 393     | 39  |
| 18.02   | 0.71 | 389     | 39  | 384     | 38  |
| 18.53   | 0.73 | 393     | 39  | 374     | 37  |



**Fig. 80** Overall photomerge view of metallography sample 1



**Fig. 84** Intergranular cracking indicative of thermal cracking. Original magnification: 200×

The wheel profiles were measured using a digital wheel profilometer and are shown in Fig. 85 and 86. No exceptions were taken with the failed wheel profile. The mate wheel profile was approaching the hollow tread limit as well as the high flange limit.

Discussion

Thermal cracks were found to be present on the wheel tread. However, the thermal cracks extended farther and at angles not generally associated with thermal cracking. The thermal crack fracture surfaces did reveal intergranular cracking, and surface hardness was measured above 50 HRC, using microhardness testing. Both results are consistent with thermal cracks and transformed tread material (tempered martensite is not permitted in wheel steel) above the allowable maximum hardness for class C wheels. The branched fatigue crack networks that propagated like stair steps initiated from the thermal cracks. The reason for the stair-stepped appearance of the fatigue cracks is not known. It is likely a by-product of internal stresses caused from rim quenching the wheel in conjunction with the texture (dendritic orientation) in the microstructure below the quenched tread.

Coupler Failures

Coupler failures are often initiated at manufacturing defects; however, some casting defects are acceptable, and the failure initiates in these locations due to high draft (tensile) or buff (compressive) in-train forces caused by train handling. The following case study demonstrates this condition.

Fracture due to fatigue cracking is also an issue with couplers and draft components. Fatigue cracks initiate from internal or external flaws, either casting or wear related, and propagate slowly until an in-train force event occurs, leading to ultimate failure. It is good practice to ascertain, if possible, any train-handling issues when investigating coupler failures.

It has been observed that couplers outside of hardness specifications generally fail within a year of installation. This occurs when the heat treatment of the couplers is improperly performed. Both extremes lead to failure due to the arduous railroad environment. When the components are below the specified hardness, the failed parts are often more highly deformed than a normal brittle failure. The components that exceed the hardness range often have smoother fracture morphology, with minimal deformation due to the brittle nature of the parts.

Casting Defects Coupler Failure

Visual Inspection

A fractured coupler is shown in Fig. 87. The coupler, an E-type, grade E coupler marked with

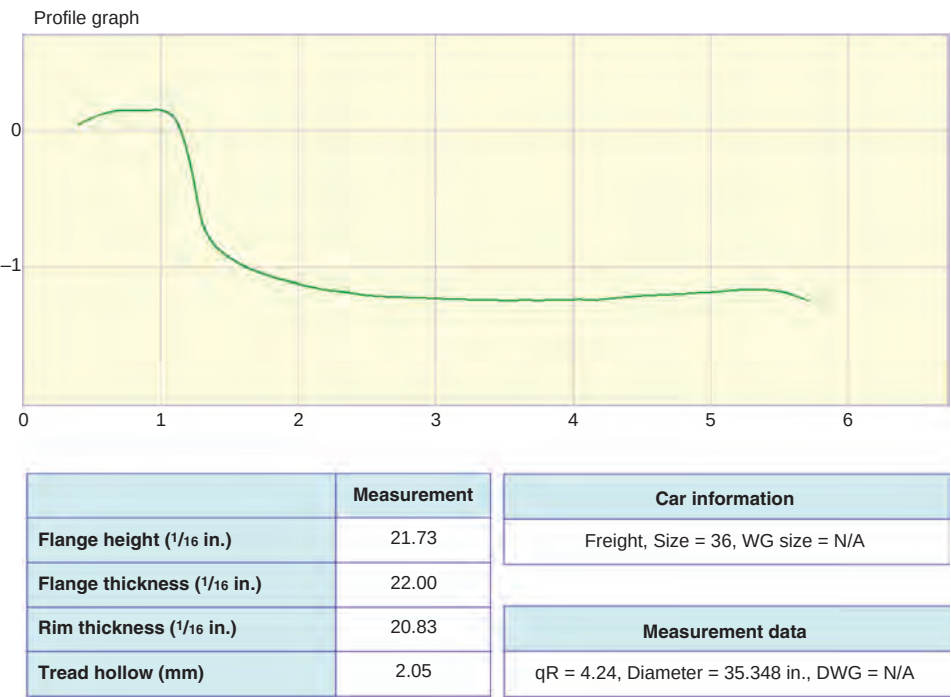


Fig. 85 Wheel profile for failed wheel NORX 2811 R1

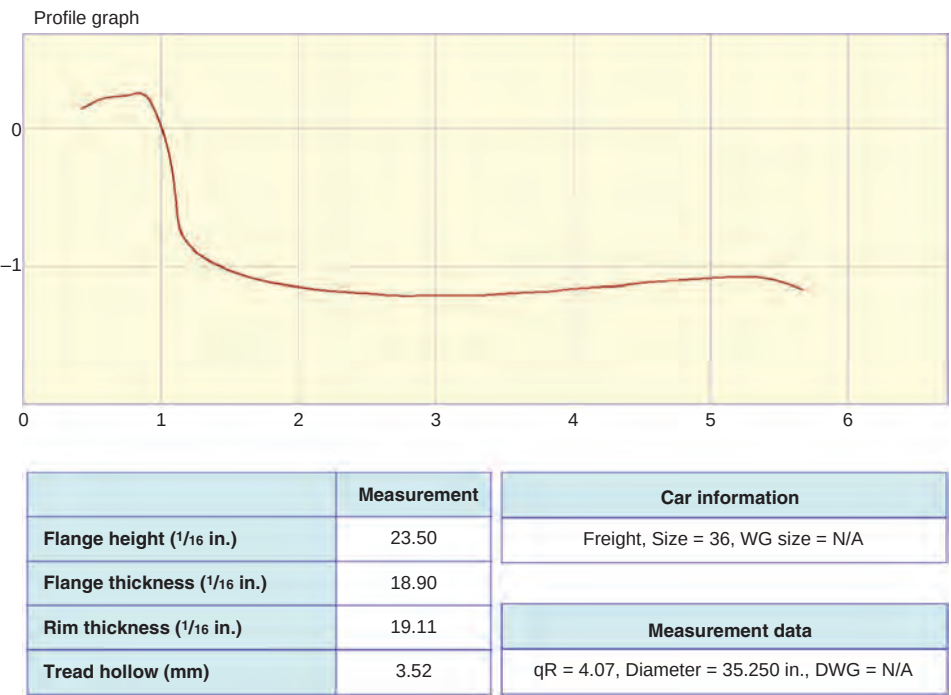


Fig. 86 Wheel profile for mate wheel NORX 2811 L1

2569, is consistent with a model number SE60EE. A section of the coupler that included the date of manufacture was missing. The coupler body was deformed, which resulted in the thrower becoming trapped within the coupler (Fig. 87).

Visual inspection of the coupler revealed a large crack behind the head (Fig. 88). The

large crack allowed the coupler body to be deformed and resulted in the thrower becoming trapped within the coupler. Figure 89 shows the coupler from a different angle, and the deformation is indicated by an arrow.

Figure 89 also shows the fracture surface, which is comprised of overload fracture with

the presence of some porosity. Figure 90 shows an initiation location that contains some porosity. Figure 91 shows the other side of the fracture surface and another initiation site. No fatigue cracking was noted on any of the fracture surfaces.

### Mechanical and Chemical Analysis

The chemical composition of the coupler was analyzed using an ARL optical emission



Fig. 87 As-received failed coupler



Fig. 88 View of cracking, which caused the thrower to be trapped



Fig. 89 Overall view of the fracture face. Arrow indicates deformation at crack shown in Fig. 88

spectrometer. The results were compared to the grade E specification of AAR M-201. The results of the analysis are shown in Table 6. The coupler material was within specifications.

A tensile test was performed on a sample to obtain mechanical properties, and the results were compared to the requirements of the grade E specification of AAR M-201. The coupler met the strength requirements but failed the ductility portion. The results of the analysis are shown in Table 7.

Brinell hardness tests were performed on a sample, and the results were compared to the requirements of the grade E specification of AAR M-201. The hardness results were within the specified hardness range. The results of the analysis are shown in Table 8.

### Discussion

Defects, in the form of porosity and solidification cracks, were present in the casting. There was no evidence of fatigue fracture emanating from the defects present. The fact that no fatigue was noted, in addition to the presence of the coupler body deformation, indicated that the coupler failed due to a singular

high-draft load event. The coupler material did exceed the strength requirements but did not meet the ductility specifications, which may have contributed to why this coupler failed, as opposed to a different component.

### Conclusions

The coupler contained casting defects in the form of porosity or solidification cracks. These defects are not uncommon in couplers and are not considered to render the coupler defective. The coupler failed due to an overload event, resulting in the sudden overload failure and deformation of the coupler body, which resulted in the thrower being trapped within the coupler. Due to the size of the defects, it is not likely that they contributed significantly to the cause of the failure. If the defects were not present, the failure likely would still have occurred. The coupler met the chemical composition and hardness requirements for AAR M-201 grade E cast steel. The coupler failed to meet the ductility requirements for AAR M-201 grade E cast steel; the reduction in area and percent elongation were below the

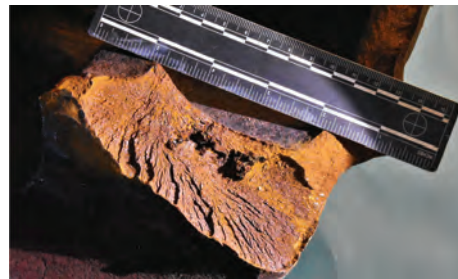


Fig. 90 Fracture surface at the bottom of Fig. 89. Some porosity is present in the fracture



Fig. 91 Fracture surface to the left in Fig. 89. Overload failure

Table 6 Chemical analysis in accordance with ASTM International Standard E415

|                   | Chemical composition, wt% |      |       |       |      |      |      |      |      |      |      |
|-------------------|---------------------------|------|-------|-------|------|------|------|------|------|------|------|
|                   | C                         | Mn   | P     | S     | Si   | Cu   | Cr   | Ni   | Mo   | V    | Al   |
| AAR M-201 grade E |                           |      |       |       |      |      |      |      |      |      |      |
| Maximum           | 0.32                      | 1.85 | 0.040 | 0.040 | 1.50 | ...  | ...  | ...  | ...  | ...  | ...  |
| Minimum           | ...                       | ...  | ...   | ...   | ...  | ...  | ...  | ...  | ...  | ...  | ...  |
| Failed coupler    | 0.28                      | 0.95 | 0.026 | 0.024 | 0.48 | 0.19 | 0.52 | 0.43 | 0.14 | 0.03 | 0.01 |

AAR, Association of American Railroads

Table 7 Tensile test results in accordance with ASTM International Standard E8

|                   | Dimensions |       | Area            |                  | Yield load (0.2% offset) |        | Yield strength (0.2% offset) |         | Ultimate tensile load |        | Ultimate tensile strength |         | Elongation after fracture (50 mm, or 2 in., gage), % |  | Reduction of area, % |
|-------------------|------------|-------|-----------------|------------------|--------------------------|--------|------------------------------|---------|-----------------------|--------|---------------------------|---------|------------------------------------------------------|--|----------------------|
|                   | mm         | in.   | mm <sup>2</sup> | in. <sup>2</sup> | N                        | lbf    | MPa                          | psi     | N                     | lbf    | MPa                       | psi     |                                                      |  |                      |
| AAR M-201 grade E | ...        | ...   | ...             | ...              | ...                      | ...    | 690                          | 100,000 | ...                   | ...    | 827                       | 120,000 | 14                                                   |  | 30                   |
| Failed coupler    | 12.7       | 0.501 | 127.2           | 0.1971           | 105,837                  | 23,793 | 834                          | 121,000 | 123,816               | 27,835 | 972                       | 141,000 | 8.3                                                  |  | 24                   |

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specifications. The coupler likely failed due to a high-draft loading event such as a runin.

Rail Failures and Rail Weld Failures

Failure Analysis of Rail Suspected of Causing Derailment

There are many failure modes of rails in the railroad industry. Frequently the question is not what the mode of failure was but rather which component failed first. For example, did the rail fail first, or was a broken wheel the cause of the derailment? If the rail failed, was there a problem in a weld, in the rail steel itself, or was it due to a lack of preventive maintenance? Derailments are destructive and often expose the weak point in rails and other components as thousands of tons of railcars violently impact. The weak points exposed are often at welds or minor defects in the rail that, in time, would be found and remedied by ultrasonic detector cars and rail crews. There are numerous failure modes of rail. The following examples are two case studies showing that what appears as an obvious cause may be, upon further investigation, a contributing factor but not the entire reason for derailment.

Background

Three sections of broken rail related to a train derailment were submitted to ascertain if the rail fracture was the cause of the derailment or if the rail failed due to the derailment. Three sections of as-received rail are shown in Fig. 92.

Table 8 Brinell hardness testing results in accordance with ASTM International Standard E18

|                   | Test readings, HBW | Comments        |
|-------------------|--------------------|-----------------|
| AAR M-201 grade E | ...                | 241–311 HBW     |
| Failed coupler    | 302, 302, 302      | 302 HBW average |

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Fig. 92 Three as-received rail sections

The two complete rail cross sections are a mating pair sharing a fracture surface. The third section is from a second rail failure location collected from the derailment. It was indicated that all three pieces were from within 6.1 m (20 ft) of the point of derailment.

Discussion

Three sections of rail were supplied. Two sections (Fig. 93) share a common fracture surface, which shows evidence of a subsurface fatigue-initiation site on the field side of the rail head (Fig. 94, 95). Although this fatigue site would have weakened the rail head, it is not the cause of the derailment, because no wheel leaving or receiving batter is associated with it. The rail fractured at this location as a result of the derailment, due in part to the weakened state due to the fatigue crack, referred to as a detail fracture in the rail industry. The third section of supplied rail contains a fracture that initiated at a weld in the rail (Fig. 96). Wheel impact damage at the fracture site is also visible in Fig. 96 and indicates that the rail was in a fractured state as freight car wheels continued passing over it. The fracture continued to grow at this site and eventually ejected the recovered fractured

rail section. The absence of this portion of the rail led to the derailment.

Conclusions

The fracture in the rail sections shown in Fig. 93 is a result of the derailment and not the cause of the derailment, despite the presence of a detail (subsurface fatigue) crack. The rail section shown in Fig. 96 fractured at the location of an in-track weld and shows evidence of wheel traffic following the fracture initiation. The rail section shown in Fig. 96 was the cause of the derailment. Based on the materials provided, the rail fractured at or near a weld during the passing of wheels, causing the derailment. While the components provided did not include the “smoking gun” section, the evidence present in Fig. 96 is consistent with a broken-rail-caused derailment.

Aluminothermic (also Known as Field or Thermite) Weld

Introduction

A fractured aluminothermic (thermite or field) weld was submitted for failure analysis. The weld (Fig. 97) was installed on August



Fig. 93 Mating fracture surfaces with thumbnail fatigue fracture from field-side corner of rail head. Black tag is a partial record of rail weld not included



Fig. 94 Subsurface fatigue fracture surface initiating from field side of rail head. No discoloration indicates the crack was not open to the atmosphere



Fig. 95 Mating fracture surface of Fig. 94. Note the damage caused by postfracture wheel impact



**Fig. 96** Fracture surface and wheel impact damage. Red arrow indicates weld material and grinding marks

13, 2019, and failed on August 14, 2019. The fracture occurred within the weld metal.

### Analysis

The rail and included weld were visually examined in the as-received state. It was noted that voids were present in the head of the rail (Fig. 98, 99). Figure 100 shows the fracture surfaces, with the initiation location circled and a void in the corner of the base indicated by an arrow.

The fracture initiated from a defect from within the web-to-base transition region. The region appeared to contain several cold laps within the weld material. A macroetch of the rail sections was performed by creating a vertical, longitudinal cross section through the centerline of the weld. The sections were ground and etched in a 50% hydrochloric, 50% water solution at approximately 70 °C (160 °F) for 20 min.

Figure 101 shows the results of the macroetch. The fracture-initiation region was within the weld metal. It was noted that the heat-affected zone was approximately 12 cm (4.75 in.) wide in the head of the rail and 8 cm (3.25 in.) wide in the base of the rail. The weld metal region was also asymmetrical (Fig. 101).

Samples were removed from the fracture-initiation region and then mounted and polished following standard metallographic procedures. The samples were examined in the as-polished state, and numerous round inclusions were present near the defect that acted as the fracture-initiation site (Fig. 102). The samples were etched with a 2% nital solution and further examined. The microstructure of the sample was consistent with pearlite (Fig. 103). The round inclusions were present along the fracture surface; secondary cracking was also present in areas along the fracture surface.

Scanning electron microscopy (SEM) was performed on a section of the fracture surface

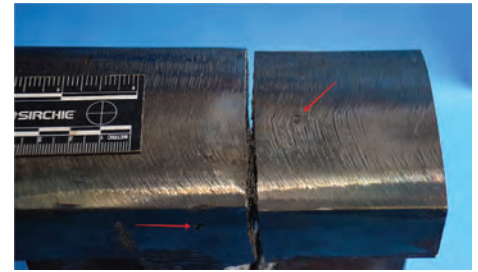


**Fig. 97** Overall view of failed weld in the as-received condition



**Fig. 99** Rail fracture surface. Circle indicates fracture initiation, and arrow indicates void

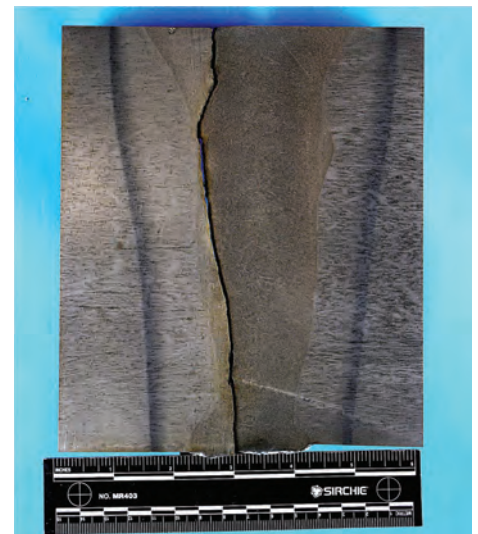
containing the initiation site. The round inclusions present in the microstructure were also present along the fracture surface (Fig. 104, 105). The areas indicated in Fig. 105 were analyzed using energy-dispersive x-ray spectroscopy (EDS). All three areas were consistent with each other. The representative spectrum is shown in Fig. 106. The inclusions consisted of mainly aluminum oxide. The fracture surface was examined using the backscatter detector of the SEM; the detector provides contrast between different elements on the fracture surface. It was noted that the defect appeared darker than the base metal (Fig. 107). An EDS analysis was performed on an area of the defect, and the spectrum was consistent with the inclusions present along the fracture face (Fig. 108).



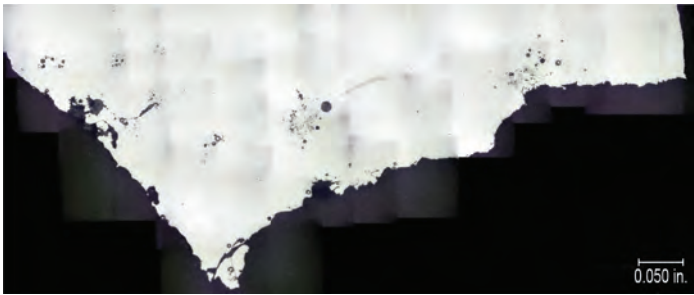
**Fig. 98** Head of rail. Arrows indicate the voids present



**Fig. 100** Mate fracture surface. Circle indicates fracture initiation, and arrow indicates void



**Fig. 101** Macroetch results of failed weld



**Fig. 102** Polished photomerge of fracture surface

**Discussion**

The thermite weld reaction consists of two main parts: the fuel (aluminum) and the oxidizer (iron oxide). The presence of aluminum-rich inclusions in the weld indicates that there was a problem with the welding process. The aluminum should be consumed during the thermite reaction and only be present as an alloying element within the rail metal. It is likely that, due to the presence of aluminum on the defect surface, the thermite reaction was still occurring when the weld charge dropped. Thus, the weld metal filled the gap; however, excessive aluminum was still present and contaminated the weld metal. The excess aluminum caused the weld to have insufficient strength, thereby causing the failure shortly after installation.

**Conclusions**

The failure initiated from a defect within the weld metal. The macroetch revealed an asymmetrical weld and heat-affected zone. Aluminum inclusions were present within the weld metal. The surface of the defect was high in aluminum content. If the aluminum is not fully consumed during the process of performing an aluminothermic weld, the weld metal will contain excessive aluminum and have substandard properties. This condition can occur if the weld metal is released too soon.

**Track Equipment Failure**

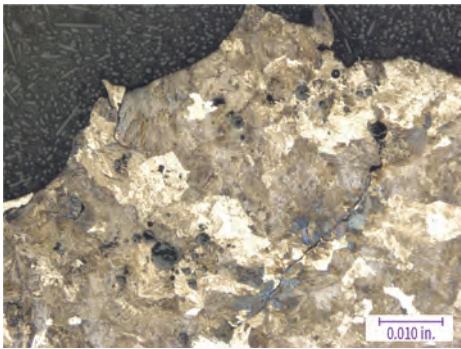
**Undercutter Chain Link**

A link from an undercutting chain fractured in service and was submitted for failure analysis. The fractured link is shown in Fig. 109.

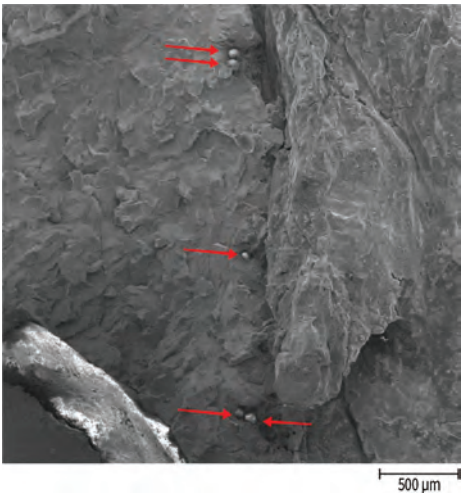
**Metallurgical Properties**

The chemical composition of the link was determined using a spark-emission spectrograph; the results are shown in Table 9. The results of the chemical analysis indicate that the link was manufactured from AISI 4140 steel.

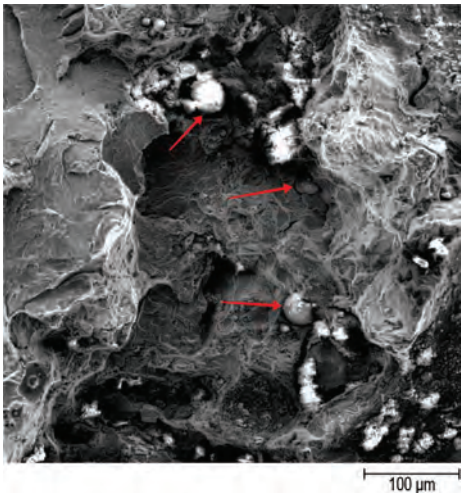
An average Brinell hardness of 285 was measured on the link section surface of the piece at right in Fig. 109. The measured hardness converts to an ultimate tensile strength of approximately 972 MPa (141,000 psi). The links were



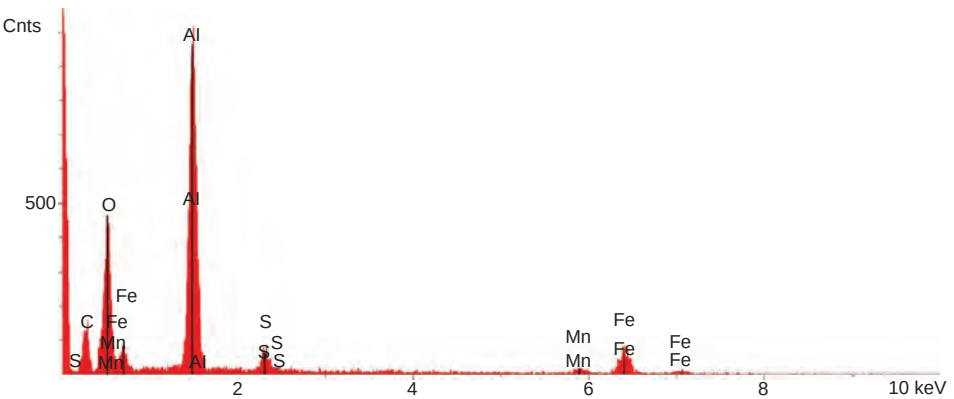
**Fig. 103** Micrograph showing inclusions and cracking near the fracture surface (top of image)



**Fig. 104** Scanning electron micrograph. Arrows indicate the presence of round inclusions. Original magnification: 55×



**Fig. 105** Closeup of fracture-surface defect. Arrows indicate spherical inclusions. Original magnification: 300×



**Fig. 106** Representative energy-dispersive x-ray spectroscopy spectrum of spherical inclusions analyzed from Fig. 105

therefore oil quenched from 840 °C (1550 °F) and tempered at 540 °C (1000 °F).

**Fracture Analysis**

The cause of the ultimate failure was a series of fatigue cracks. Some of the cracks grew to depths of up to 25 mm (1 in.) before sudden rupture occurred. Figure 110 shows

that fatigue cracks (identified by the clam-shell progressive crack growth) occurred at multiple locations.

Metallographic analysis revealed that the quenching and tempering heat treatment of the links produced a bainitic microstructure, which is expected in a quenched-and-tempered 4140 steel. However, small intergranular cracks at the chain surface indicate that

quench cracks had developed during heat treating (Fig. 111).

A heavy layer of oxidation was present on a portion of the fracture surface (Fig. 110). The



**Fig. 107** Backscatter image. Left side of image is base metal; right side is the defect area. Original magnification: 116×

oxide surface prevented inspection of the fracture surface and was therefore removed, using an inhibited hydrochloric and dibutyl thiourea solution. Examination of the surface cracking by SEM verified the presence of intergranular fracture (Fig. 112).

### Retarder Brake Beam

A retarder beam was removed from service due to an observed crack during daily inspection. A retarder is used as an external brake applied to the front and back rim faces of a wheel to slow railcars in hump yards, while using gravity to sort individual cars down a hill through switches to one of many tracks during train assembly. A metallurgical failure analysis was requested to determine if the cracking occurred due to a manufacturing defect or if improper repair processes had been employed. In Fig. 113 and 114, the beam is shown in the as-received condition, with a crack outlined in chalk.

The beam was photographed and visually inspected to ascertain the best method for opening the crack. This was achieved by cutting the back side of the beam. The fracture

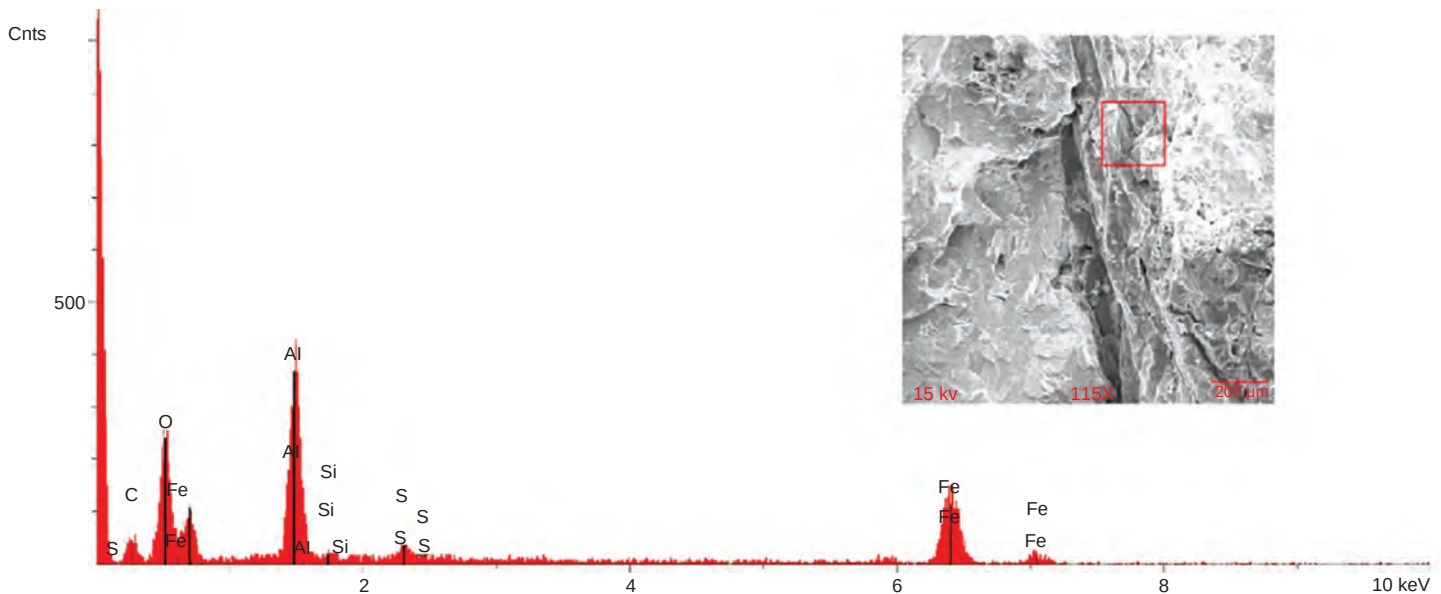
surfaces were exposed (Fig. 115). The morphology of the fracture surfaces was interpreted as typical fatigue fracture, as indicated by the beach marks present. The arrows are oriented toward the crack-initiation location (Fig. 116).

After opening the crack to reveal the surfaces, several additional cuts were performed. This cutting provided samples that were used to obtain chemical composition, mechanical properties, and for metallography and SEM of the fracture surfaces. The sectioning locations for the metallographic and SEM samples are shown in Fig. 117.

The microstructure sample was mounted and polished according to standard metallographic procedures. The sample was etched with a 2% nital reagent and inspected using a digital microscope. The microstructure was consistent with tempered martensite, which is expected for a quenched-and-tempered steel (Fig. 118).

The SEM did not reveal any defects or inclusions within the base material at the location of the fracture surface. The fracture surface had heavy oxidation present, and efforts to clean the surface were marginal, thus obscuring areas of the fracture surface.

Chemical composition was performed on a sample of the retarder beam. The chemical



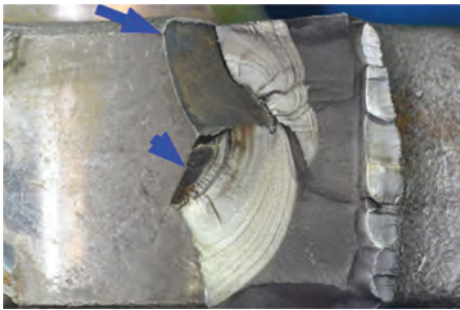
**Fig. 108** Energy-dispersive x-ray spectroscopy spectrum from an area of the defect shown in Fig. 107



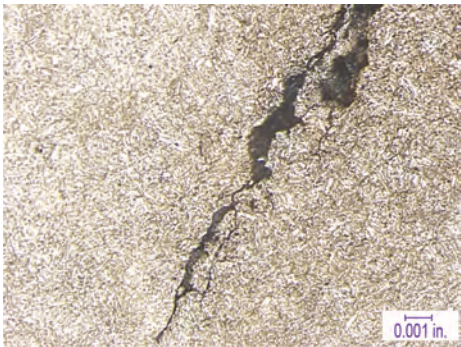
**Fig. 109** Fractured undercutter chain link in the as-received condition

**Table 9** Chemical composition of undercutter link

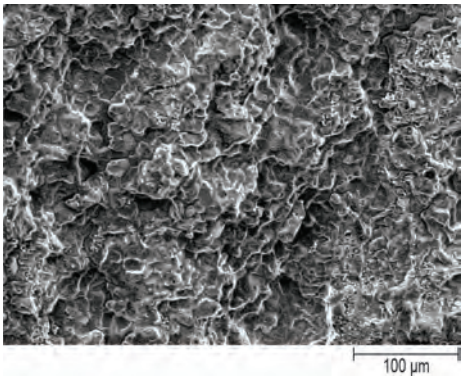
|            | Chemical composition, wt% |           |           |           |       |           |           |       |       |
|------------|---------------------------|-----------|-----------|-----------|-------|-----------|-----------|-------|-------|
|            | C                         | Mn        | P         | S         | Ni    | Cr        | Mo        | V     | Al    |
| Chain      | 0.40                      | 0.67      | 0.001     | 0.012     | 0.018 | 1.11      | 0.18      | 0.007 | 0.019 |
| Grade 4140 | 0.40–0.45                 | 0.75–1.00 | 0.040 max | 0.040 max | ...   | 0.80–1.10 | 0.15–0.25 | ...   | ...   |



**Fig. 110** Multiple fatigue cracks in chain link. Arrows indicate fracture surfaces with heavy oxide



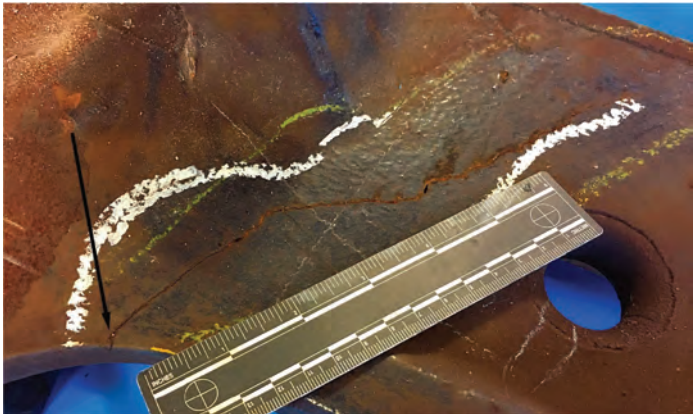
**Fig. 111** Microstructure of chain with intergranular surface crack. Nital etch. Original magnification: 1000×



**Fig. 112** Scanning electron microscope image of surface crack indicating intergranular fracture. Original magnification: 350×



**Fig. 113** Overall view of as-received beam. Crack location is indicated by chalk and arrow on right side of image



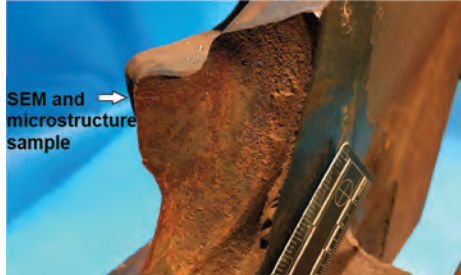
**Fig. 114** Closer view of crack in beam. Arrow indicates the general area of initiation



**Fig. 115** Fracture surface opened. Black arrows correspond to crack-initiation sites on the mating fracture surfaces



**Fig. 116** Close-up view of fatigue fracture surface with beach marks. Arrows indicate the initiation-site areas



**Fig. 117** Sectioning plane (dashed line) through initiation area. SEM, scanning electron microscopy

composition is shown in Table 10, along with the specification limits. Table 10 indicates that the carbon content is out of specification, which causes the carbon equivalent to be out of specification. The elevated carbon would make the steel more hardenable; however, when comparing the results of the mechanical testing (Table 11), the material is within mechanical property specifications for strength

but just misses the minimum elongation and reduction of area.

### Discussion

The elevated carbon content would lead to higher strength and reduced ductility, as seen in the low elongation and reduction-of-area values. Consequently, the casting would be more brittle. Fracture toughness and fatigue

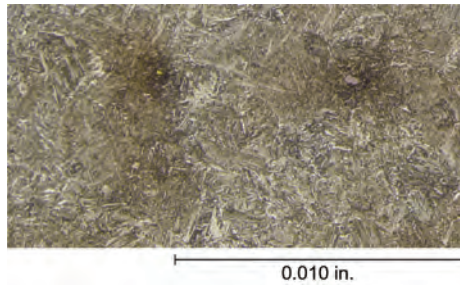


Fig. 118 Microstructure is tempered martensite

Table 10 Chemical composition of retarder and specification limits from customer standard

|                   | Chemical composition, wt% |      |           |           |          |      |      |      |      |      |                   |
|-------------------|---------------------------|------|-----------|-----------|----------|------|------|------|------|------|-------------------|
|                   | C                         | Mn   | P         | S         | Si       | Cr   | Ni   | Mo   | Cu   | V    | Carbon equivalent |
| Customer standard |                           |      |           |           |          |      |      |      |      |      |                   |
| Maximum           | 0.33                      | 1.0  | 0.050 max | 0.025 max | 0.60 max | 0.65 | 0.80 | 0.25 | ...  | ...  | 0.72              |
| Minimum           | 0.28                      | 0.75 | ...       | ...       | ...      | 0.40 | 0.50 | 0.20 | ...  | ...  | 0.56              |
| Sample            | 0.44                      | 0.81 | 0.019     | 0.019     | 0.75     | 0.57 | 0.52 | 0.20 | 0.18 | 0.01 | 0.78(a)           |

(a) Carbon equivalent =  $w_C + w_{Mn}6 + (w_{Cr} + w_{Mo} + w_V)5 + (w_{Ni} + w_{Cu})15$ , where  $w_i$  is the weight percent of alloying element  $i$

Table 11 Mechanical property results

|               | Dimensions |       | Area            |                  | Yield load(0.2% offset) |        | Yield strength(0.2% offset) |            | Ultimate tensile load |        | Ultimate tensile strength |             | Elongation after fracture<br>(50 mm, or 2 in., gage), % | Reduction of area, % |
|---------------|------------|-------|-----------------|------------------|-------------------------|--------|-----------------------------|------------|-----------------------|--------|---------------------------|-------------|---------------------------------------------------------|----------------------|
|               | mm         | in.   | mm <sup>2</sup> | in. <sup>2</sup> | N                       | lbf    | MPa                         | psi        | N                     | lbf    | MPa                       | psi         |                                                         |                      |
| Sample 1      | 12.7       | 0.500 | 127             | 0.1963           | 95,160                  | 21,393 | 752                         | 109,000    | 118,452               | 26,629 | 938                       | 136,000     | 16                                                      | 32                   |
| Specification | ...        | ...   | ...             | ...              | ...                     | ...    | 586                         | 85,000 min | ...                   | ...    | 724                       | 105,000 min | 17 min                                                  | 35 min               |

evaluations were beyond the scope of the analysis. The initiation site of the crack did not contain any abnormal stress concentrations (such as repairs, nicks, or gouges). Some locations on the beam suggested that wear conditions were present. This is likely normal but could indicate either an accelerated condition due to excessive forces or possibly a misalignment. With the available beam, it is not possible to be more definitive.

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