

Handbook of Materials Failure Analysis

With Case Studies from the Chemical,
Concrete, and Power Industries



Edited by
Abdel Salam Hamdy Makhlouf
Mahmood Aliofkhazraei



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Preface

Failure in different industries can happen in segments of the machines when dynamic loading occurs. When you think about fatigue, you may think of an aerial ladder, the crankshaft of an automobile, or the wing of a plane. For failure analysis of the segments structure, we need mechanical examinations. For example, to predict the lifetime of a spring of a garage door, it should be clear that how much load is applied to the spring. Also, it should be clear how many times this force is applied, for example over a period of one year, on the spring. Using analytical methods and computer modeling methods, it can be predicted after how many cycles the fatigue phenomenon will occur in the spring. After comparison between the results of the fatigue test and the computer modeling, the validity of the model can be investigated. Using fractography, it can be made clear whether or not a fracture happened due to the fatigue. Fractography is the act of investigating the fracture surface, using a microscope.

This handbook provides an in-depth examination of materials failure in specific situations, a vital component in both developing and engineering new solutions. It covers analysis of materials failure in the areas of chemicals, power, and structures, wherein the failure of a single component can result in devastating consequences and costs. Material defects, mechanical failure as a result of improper design, corrosion, surface fracture, and other failure mechanisms, are described in the context of real-world case studies involving steam generators, boiler tubes, gas turbine blades, welded structures, chemical conversion reactors, and more. This book is an indispensable reference for engineers and scientists studying the mechanisms of failure in these fields.

This handbook contains many real-world failure cases and case studies covering a wide spectrum of materials failure related to chemicals, concrete, and power applications. The editors thank all the contributors for their excellent chapter contributions to this handbook, and for their hard work and patience during preparation and production of the handbook. We sincerely hope that the publication of this handbook will help people from Industry and Academia to get the maximum benefits from the experience contained in the published chapters.

Summer 2015
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Progressive failures of components in chemical process industry: Case history investigation and root-cause analysis

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1 OVERVIEW OF FAILURE MECHANISMS

Failure analysis is a structured procedure that can be implemented using a process-based approach aiming to determine the damage mechanisms, discover the root-cause(s), and outline the corrective actions, which are necessary precursors for the improvement of a system [1]. This is an approach common in almost every

technical/economic activity offering significant potential in problem-solving viable growth.

Metal components exhibit lower service life when operated at elevated temperatures. Depending on the particular environmental and loading conditions, the most common causes of elevated-temperature failures can be classified as creep-originated, environmentally induced fractures, high-temperature fatigue, and thermal fatigue [2]. Creep is realized through the nucleation and growth of transgranular or intergranular voids depended on the applied operating conditions, that is, stress and temperature [3]. Creep is a progressive time-dependent deformation which leads to final rupture with potential catastrophic consequence and implications to health, safety, and environment. Remaining life-assessment techniques are a valuable asset in failure prevention and cost minimization of high-temperature working equipment. Methods using the Larson-Miller parameter (LMP), expressed in Equation 1.1, life fraction rules, and damage accumulation conditions are very popular in remaining life prediction [4,5], along with finite-element analysis (FEA) modeling presented in the relevant literature [6]

$$\text{LMP} = T \times (23 + \log t) \quad (1.1)$$

where T is the operating temperature (K) and t is the service time (h).

Larson-Miller creep rupture curves present the stress variation as a function of LMP. Knowledge of stress and temperature conditions could therefore lead to the determination of the estimated creep life with at a certain statistical confidence level. Life fraction is used for the estimation of remaining life of structural elements exposed at high temperatures, according to Equation 1.2:

$$\frac{t_{\text{op}}}{T_{\text{op}}} + \frac{t_{\text{test}}}{T_{\text{test}}} = 1 \quad (1.2)$$

where t_{op} is the operating time, T_{op} is the total operating time at service conditions, t_{test} is the time for rupture at the test under accelerating conditions, and T_{test} is the time for rupture of fresh component under accelerating conditions.

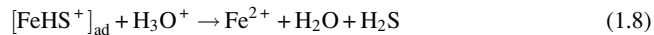
The commonly referred eight forms of corrosion were addressed in the classical corrosion literature [7]. The issue of stress corrosion cracking (SCC) of stainless steels is a critical research field due to the extensive use of such materials in special applications and the associated risks (unexpected failure without warning, severe impact to human safety and environment). It is a dominant failure mechanism for metallic components in unit operations industries that embrace many types of physical and chemical processes, under elevated temperatures, high pressures, and chemically reactive environments. SCC of a stainless steel spiral heat exchanger in a distillation column has been attributed to the action of chloride ions in the cooling water, compromising the integrity of the protective oxide layer, producing corrosion pits and facilitating crack formation [8]. SCC was also attributed to the detrimental effect of chloride ions as reported in the case of failure of an AISI 304 styrene

seashore storage tank [9] and AISI 304 ethylene glycol condensate collector vessel [10]. In case of oil refineries, SCC is very frequently encountered and has also been correlated to the presence of sulfur-containing compounds. Specific examples involve the effect of H₂S on failure of SS347H tubes in a hydroprocessing reactor heater [11] and of polythionic acid on the failure of AISI 321 hydrocracker heat exchanger tubes [12]. SCC was also the dominant failure mechanism in a ring-type component of hydrocracker unit [13]. The effect of sulfur compounds has been reported to be of significant importance, especially in the case of naphtha feedstock circulation, due to the presence of H₂S in these fluids in relatively high concentration. In Ref. [14], transgranular cracking and intergranular SCC were noticed in case of naphtha hydrotreater furnace tubes. The influence of corrosion environment on fatigue-related processes is reviewed in Ref. [15]. The damaging role of corrosive species, such as Cl and S, is underlined also in case of localized corrosion of steel piping elbow in oil-gas separation system initiated from inorganic salts in crude oil [16], as outlined by the chemical reactions 1.3–1.5.



Hydrolysis of inorganic salts produces HCl (chemical reaction 1.3). The dissolution of HCl in water creates aggressive corrosion conditions, leading to steel chemical attack (chemical reaction 1.4). Under the presence of H₂S, regeneration of HCl is achieved, sustaining the corrosive conditions (chemical reaction 1.5). FeS can produce more stable FeS₂ reacting with elemental S, which could have been produced by bacterial activity and the cycle is continued by the regeneration of fresh HCl. Metal loss is mainly caused by HCl acid attack, produced by hydrolysis of inorganic chloride salts, such as CaCl₂ [16].

The detrimental effect of the presence of H₂S aqueous solution was also addressed in a case of steel absorber corrosion in a desulfurization plant, by the following chemical reactions [17]:



where “ad” denotes adsorbed substance.

In petroleum and chemical processing industry, gas transportation systems suffer from CO₂ corrosion manifested by the formation of a protective scale consisting mainly of FeCO₃, once Fe²⁺ and CO₃²⁻ concentrations are above solubility limit [18]:



FeCO₃ scale constitutes a diffusion barrier which impedes ion transfer and minimizes steel surface exposure, hindering therefore the evolution of corrosion reactions.

The presence of hydrogen sulfide (H₂S) in gas and oil industries induces the formation of atomic hydrogen which diffused-in and trapped-in atomic scale defects, dislocations and nonmetallic inclusion sites leading to hydrogen-induced cracking (HIC) incidents in pipelines [19]. Although HIC-type failure is a subject of major importance in relevant industries and numerous studies are focused on this topic, it is out of the scope of this chapter.

A great variety of failure mechanisms is related to chemical process industries, some of them are summarized in Table 1.1.

2 ANALYTICAL TECHNIQUES

Macroscopic observations of the failed components were performed with a high-definition camera and a stereomicroscope. In order to reveal the main failure mechanisms, microscopic observations were carried out using an SFEG scanning electron microscope (SEM) coupled with an energy dispersive X-ray spectrometry (EDS) detector for elemental analysis of selected areas. Metallographic examination was conducted using an inverted optical microscope. Prior to that, cross-sections of the failed components were prepared using hot-mounting, wet grinding up to 1200 grit SiC paper and polishing with diamond and silica suspensions. Immersion etching was performed to plain carbon steel sections using Nital 2% reagent. Hardness testing was performed on polished sections using Vickers indentation technique with a 10 kg applied load.

3 CASE STUDIES

3.1 CREEP FRACTURE OF A CARBON STEEL SUPERHEATER TUBE INSTALLED IN A POWER-PLANT UNIT

3.1.1 Summary

The present case study is focused on the metallurgical investigation pertaining to the high-temperature-induced failure of a pipeline after short-time service [20]. The pipeline circuit was utilized as superheater in a lignite power-plant unit and consisted

Table 1.1 Review of Characteristic Failure Mechanisms Activated in Chemical Process Environments Together with Some Relevant References

No.	Main Failure Mechanism	Environmental Condition(s)	Failure Mode (Typical Failure “Signature”)	Some Relevant References (Addressing the Failure Mechanism)
1	Creep damage	Long-term heating, stress (internal pressure)	Spheroidization of carbides/void formation, growth and coalescence	[4,5,20–22]
2	Thermal fatigue and high-temperature fatigue	Thermal cycling and/or high-temperature stress cycling	Macroscopically brittle fracture—progressive crack markings—oxidized surface	[2,23]
3	Hot oxidation/corrosion/erosion corrosion	Chemical attack (e.g., chlorides, sulfides, CO ₂)/high-temperature oxidation	Oxidation and corrosion scale deposits/metal loss or dissolution	[16–18]
4	Hydrogen induced cracking (HIC)	Chemical attack by H ₂ S, H ₂ , atomic hydrogen diffusion	Brittle cracking, microcracking at areas of entrapped nonmetallic inclusions	[19]
5	Stress-corrosion cracking (SCC)	Chemical attack (e.g., chlorides, sulfides) and residual or applied tensile stress	Delayed brittle fracture/intergranular or transgranular cracking	[8,9,12,13,24]
6	Corrosion fatigue (CF)	Chemical attack/pitting—cyclic loading	Surface pitting and fatigue crack nucleation	[15]
7	Microstructural degradation, sensitization, and embrittlement	Thermally induced degradation and phase transformations due to long-term overheating (~600–900 °C)	Cr-depletion, carbide precipitation and grain boundary sensitization, σ -phase formation	[24–27]

of series of steel tubes welded circumferentially. Systematic data collection over 1 year of service indicated that failures occurred approximately every 15 days on average and situated either at curved areas of the tubes or in the vicinity of circumferential welded joints. The detailed metallurgical examination followed on failed tubes indicated that pipeline damage caused as a consequence of creep rupture.

In all cases examined, the material failed due to damage evolution indicated by the rapid growth and coalescence of creep cavities, initiated at the outer tube surface and at service temperature equal to approximately 45% of the steel melting point (absolute temperature units).

3.1.2 Background information

The present research concerns the investigation of the systematic short-time service failure of a superheater pipeline circuit, part of the principal steam heat exchanger assembly, in a thermoelectrical power-plant unit located in Northern Greece. The almost 30-year old pipeline was constructed over a lignite burner. According to the original design, the whole pipeline circuit consists of a series of individual tubes of dimensions: outer diameter ~ 38 mm, wall thickness ~ 6 mm, and length ~ 4 m. These tubes are properly welded to form a pipeline with a total heat transfer area of 5200 m^2 . The proposed materials of construction were the following steel grades: 15Mo3, 13MoCr44, and 10CrMo910 (Table 1.2), commonly used for the production of pressure vessels and pipes.

The average steam linear velocity along the pipeline was $\sim 9 \text{ m s}^{-1}$ and the steam outlet flow rate is 900 t h^{-1} . The main operating conditions of the pipeline are listed in Table 1.3 [20].

An increased maintenance frequency was noticed over the past several years, reaching to high downtime and minimizing continuous operation to 15 days average.

Table 1.2 Chemical Composition of the Steel Grades (wt%), Regularly Used as Materials of Construction

Steel Alloy	W.Nr.	C	Si	Mn	P	S	Cr	Mo
10CrMo910	1.7380	0.06-0.15	≤ 0.50	0.40-0.70	≤ 0.035	≤ 0.030	2.0-2.5	0.90-1.10
15Mo3	1.5415	0.12-0.20	0.10-0.35	0.40-0.90	≤ 0.035	≤ 0.030	–	0.25-0.35
13CrMo44	1.7335	0.08-0.18	0.10-0.35	0.40-1.0	≤ 0.035	≤ 0.030	0.70-1.10	0.40-0.60

Table 1.3 Designed and Real Principal Process Parameters of Superheater Pipelines, as Approximate Values [20]

Process Parameter	Designed	Real
Steam inlet pressure (bar)	190	–
Steam outlet pressure (bar)	180	–
Steam inlet temperature ($^{\circ}\text{C}$)	370	450
Steam outlet temperature ($^{\circ}\text{C}$)	440	500

This situation resulted in high unit operation cost, due to frequent replacements and lower productivity. A failure analysis study was implemented in order to detect the failure mechanisms and causes and implement corrective actions to solve the problem.

3.1.3 Failure analysis

The principal elevated-temperature failure mechanisms that might have influenced the short lifetime of the pipeline are the following:

- Creep deformation, due to abnormal operating conditions.
- High-temperature oxidation, due to interaction with the combustion gases, the chemical composition of which can be varied according to the lignite's mineralogical structure.
- Thermal fatigue, due to the cyclic loading conditions imposed by the frequent operation interruptions for repair and maintenance of the pipeline.
- Elevated-temperature fatigue, due to cyclic variations of the steam pressure.

Metallurgical investigation of representative failed tube parts collected from the damaged pipeline (Figures 1.1 and 1.2) is discussed below.

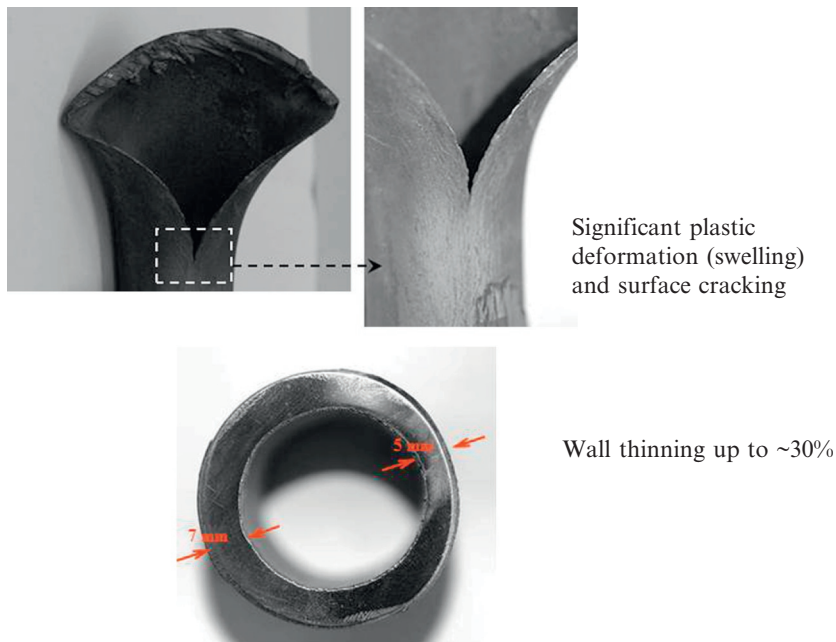
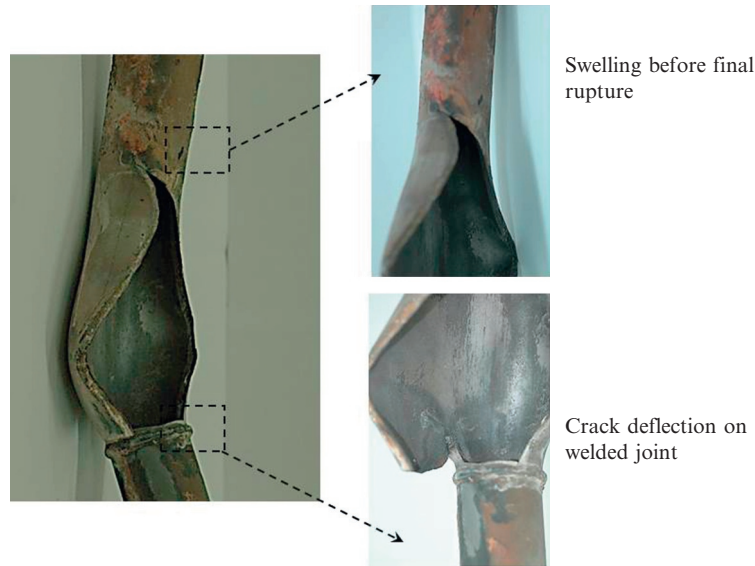


FIGURE 1.1

Macrographs of the failed tube recovered from the damaged pipeline area, showing “fish-mouth”-type rupture. Note the exaggerated swelling action and appreciable ~30% wall thinning (see also Ref. [20]).

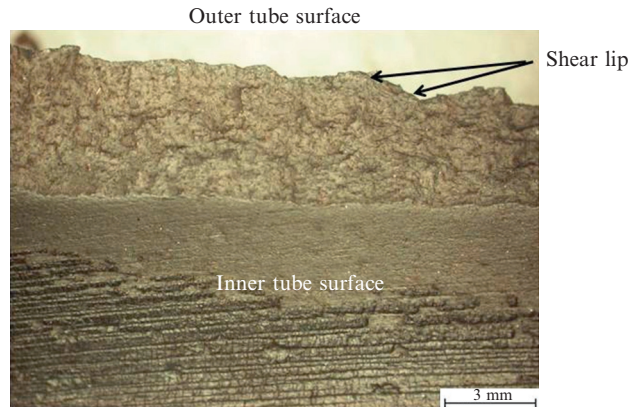
**FIGURE 1.2**

Macrograph of the curved area of the ruptured pipe, showing significant swelling and crack deflection. At the bottom image, cracking occurred at the weld joint area is noticed (see also Ref. [20]).

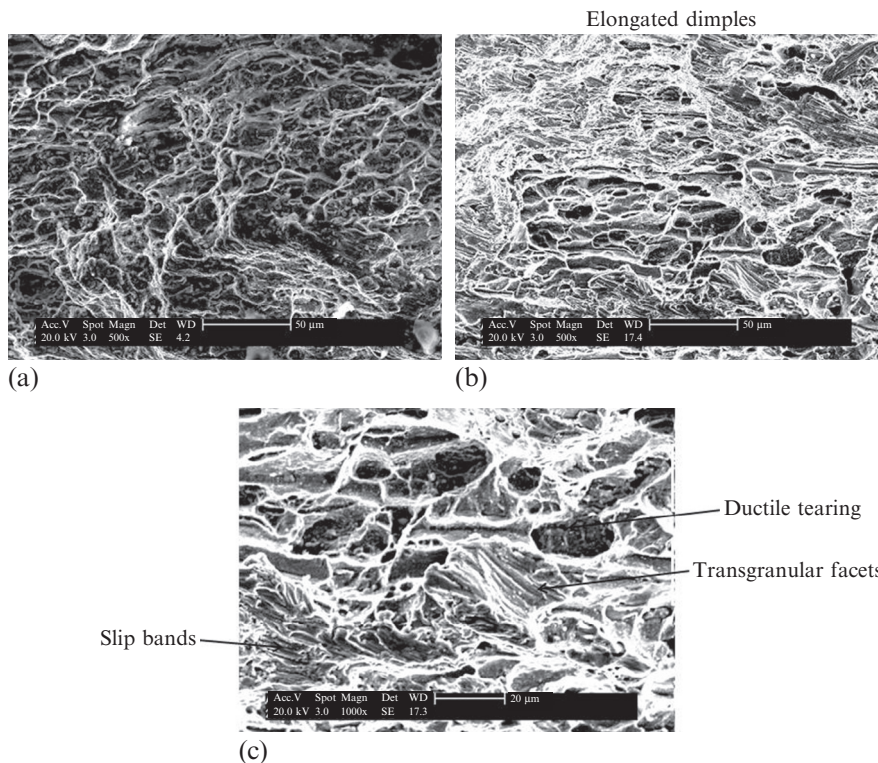
3.1.3.1 Failure at the bent areas of the pipeline

The failed parts recovered from bent tube areas exhibited the typical “fish-mouth” thin-lip rupture (Figure 1.1a), macroscopic signatures of creep-induced failure of superheater and reformer tubes [23,28]. Fracture occurred at the inner side of the bent tube. The wall thickness at the tip of the “fish-mouth” was markedly reduced from its nominal value of 6 mm down to about 1 mm. On the oxide layer, surface tearing indicated by intermittent secondary rupture cracks in the direction of the tube axis can be identified. The secondary rupture cracks appeared as slight separations at areas where the base metal is exposed due to detachment of the broken oxide layer, most likely caused during thermal stress and subsequent overhaul operations. Similar features have been observed in the case of creep failures of turbine blades reported in Ref. [29]. Substantial wall thinning (from 7 to 5 mm) extending to a distance of about 120 mm from the ruptured area corresponds to a 33% reduction of its original nominal dimension (Figure 1.1). Fractographic observations of the ruptured lips, performed by optical stereomicroscopy and SEM, are depicted in Figures 1.3 and 1.4. Characteristic features and patterns demonstrated by the fracture area clearly indicated the occurrence of ductile fracture undergone after significant prior plastic deformation. On the dimpled fracture surface (Figure 1.4), several characteristic morphologies can be readily distinguished:

- Large equiaxed voids developed from the growth and coalescence of creep cavities of smaller dimensions (Figure 1.4a).

**FIGURE 1.3**

Optical stereomicrograph depicting the fracture surface of the ruptured tube.

**FIGURE 1.4**

SEM micrographs (SE imaging) showing details of the fracture surface illustrated in Figure 1.3. Cracking progressed by microvoid coalescence leading to ductile dimples covering almost the entire fracture surface; slip bands were evident in the interior of dimples while isolated transgranular facets were also identified. Elongated dimples were also detected as a result of the fibered grain structure formed by the manufacturing process.

- Elongated dimples corresponding to the initial grain alignment due to the directionality of the manufacturing process (Figure 1.4b).
- Isolated areas of transgranular cleavage fracture and slip bands appeared in the interior of the dimples (Figure 1.4c).

The evolution of creep cavitation as an indicator of damage evolution, across the tube wall, is illustrated in Figures 1.5 and 1.6. At the outer surface of the tube, exaggerated wedge-void coalescence (Figures 1.5a and 1.6a) resulted in the genesis of intergranular surface cracks (depth $\sim 50\text{ }\mu\text{m}$) filled with hot oxidation products. Several creep void evolution mechanisms were identified, ranging from intragranular to intergranular cavitation (Figures 1.5b and c and 1.6b and d), see also Ref. [21]. The existence on the same tube cross-section of the entire creep cavitation evolution period (initiation $\rightarrow$ growth $\rightarrow$ coalescence) is suggestive of the thermal gradient established across the tube wall, which constitutes driving force for heat flux from the hot lignite combustion gases on the outer surface of the tube, to the high-pressure steam flowing inside the tube.

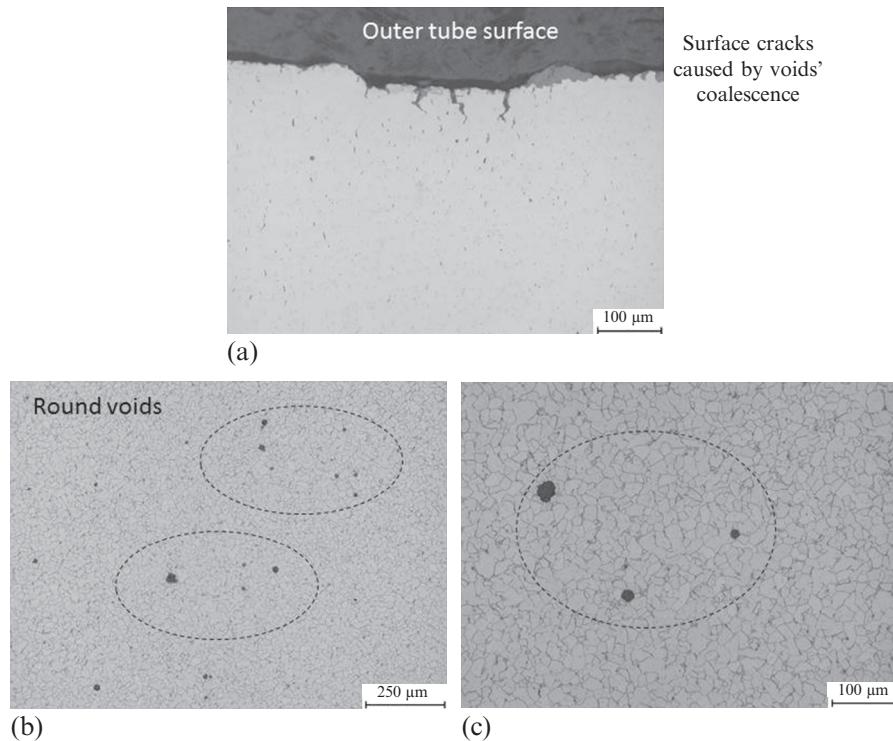
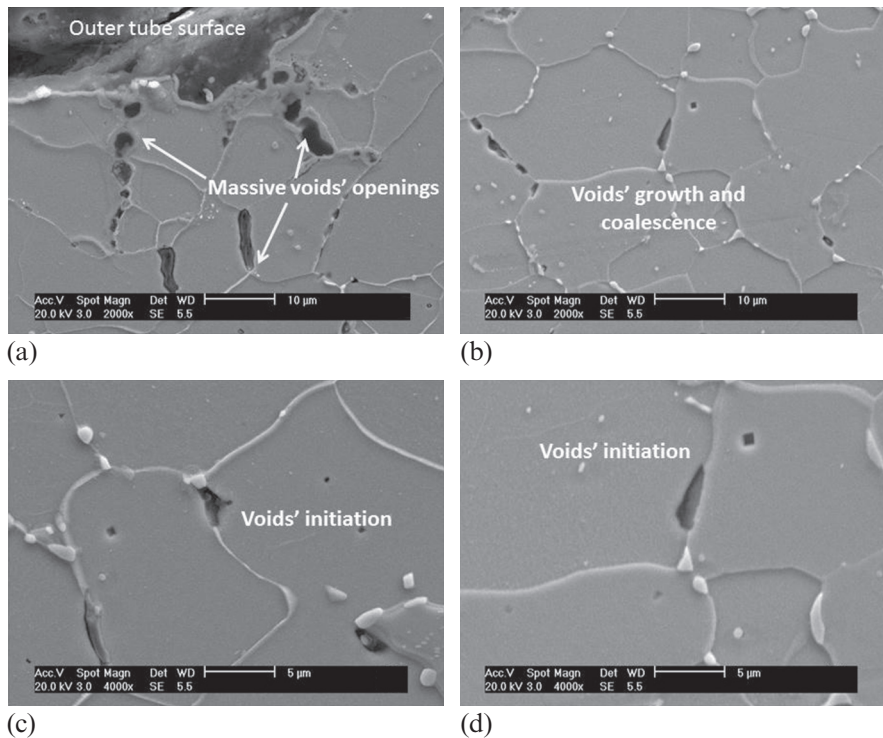


FIGURE 1.5

Optical micrographs of transverse sections close to the crack tip (see Figure 1.1), showing the evolution of creep voids: (a) close to the outer tube surface, (b) round voids growth in the wall body, and (c) detail of (b).

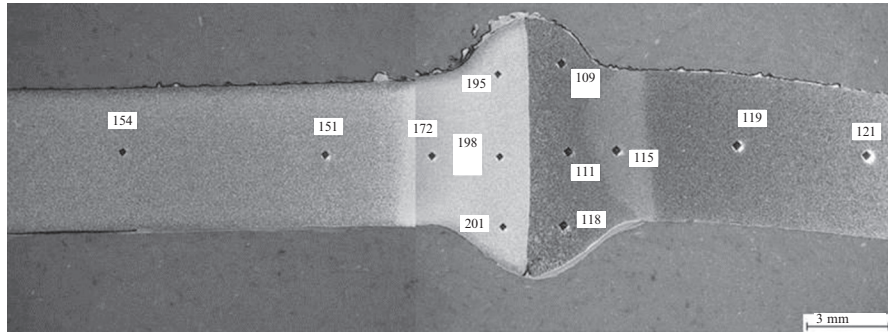
**FIGURE 1.6**

SEM micrographs outlining the progressive evolution of creep voids in the tube surface and in the interior of the tube wall body.

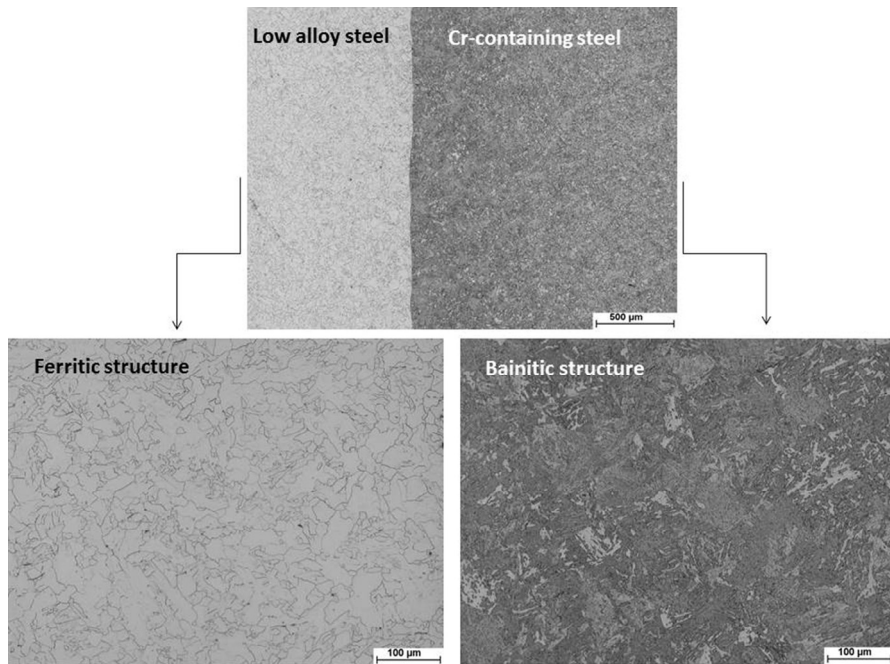
The tube material characterization was assessed by applying both metallographic and microanalytical techniques (EDS), which suggested that it was a low-alloy ferritic carbon steel, probably corresponding to the 15Mo3 steel grade commonly used for this type of application. Precipitates observed at the grain boundaries (Figure 1.6c and d) indicated cementite spheroidization and grain boundary precipitation during prolonged high-temperature regime, close to A_{c1} .

3.1.3.2 Failure at the vicinity of the welded joints

The failed parts of the pipeline removed from areas near circumferential welded joints, shown in Figure 1.2, exhibited a “cobra”-type pattern [23], with cracks initiating at the vicinity of the joint and propagating almost parallel to longitudinal axial direction. Stereomicroscopic evaluation of the welded joint, showing also hardness readings' distribution on both sides of the welded tubes, is presented in Figure 1.7. Higher magnification metallographic observations revealed both the dissimilar materials of the joined tubes, as well as the absence of fusion zone or filler material, which likely indicate the application of a high-temperature solid-state welding technique (Figure 1.8). EDS microanalysis indicated that one of the tubes was a

**FIGURE 1.7**

Optical stereomicrograph of a longitudinal section, perpendicular to the circumferential weld joint, revealing solid-state welding of dissimilar steel grades (section taken from the tube sample depicted in [Figure 1.2](#)). Note the hardness variation across the joint (Nital 2% immersion etching) (see also Ref. [20]).

**FIGURE 1.8**

Optical micrographs showing the different microstructures from both sides of the weld joint.

Cr-containing steel, probably corresponding to the 13CrMo44 or 10CrMo910 grades also normally used for such kind of application. The other tube, within which cracks have been formed, was identified as a low-carbon-low-alloy ferritic steel, as in the previous case. A clear interface is noticed between the two welded tubes; the joint seems free of flaws and discontinuities and failure is located outside the bonding line. Concerning hardness distribution on both sides of the welded joint, the following findings could be reported:

- In the case of Cr-containing carbon steel grade, the area near the seam exhibited an average hardness up to ~ 200 HV10, while at a distance of 3.5 mm from the bonding line the hardness decreased to ~ 170 HV10 toward the parent metal hardness (~ 150 HV10).
- In the case of ferritic carbon steel, no appreciable differences were accounted, that is, the hardness varied between 110 and 120 HV10 at all the examined areas, from weld zone to parent metal.

The area close to the joint boundary of Cr-containing carbon steel manifested the characteristic upper-bainite microstructure (Figure 1.9). The adjacent area of intermediate hardness exhibited the presence of tempered ferrite with a fine distribution of spheroidized cementite, while the parent metal exhibited a microstructure of ferrite with partially spheroidized cementite.

The adjacent area of ferritic low carbon steel grade consisted of equiaxed ferrite grains, of uniform distribution, an observation indicating that the material was recrystallized while the carbide precipitates found at the grain boundaries probably attributed to spheroidized cementite. In the same area, individual creep voids of small size were scarcely observed.

The parent material exhibited round and transgranular creep cavities with an average diameter of 30–50 μm (Figure 1.10). A close view on this area revealed that

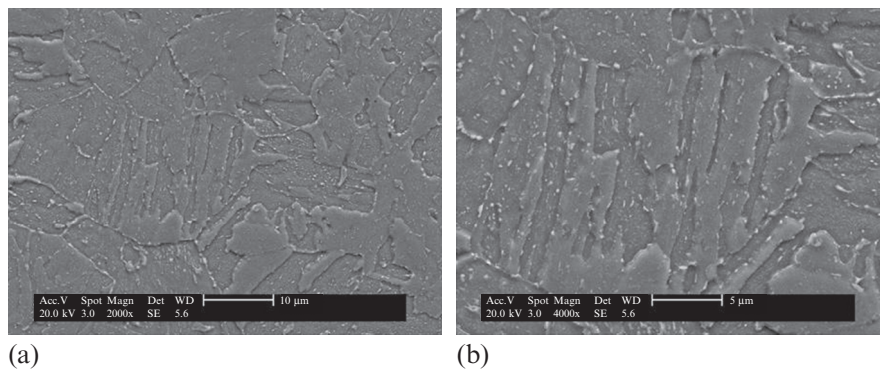
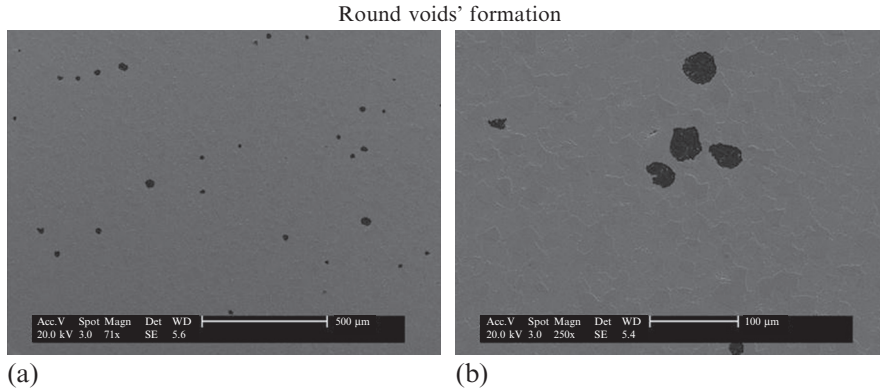


FIGURE 1.9

SEM micrographs showing details of the bainitic structure of the Cr-containing steel grade. (b) Detail of (a).

**FIGURE 1.10**

(a) SEM micrographs showing round voids' growth of the side of the ferritic low-carbon-low-alloy steel. (b) SEM micrograph showing imminent coalescence of adjacent voids and in-between formation of localized necking.

the growth and coalescence of creep voids were taking place through micronecking process between two adjacent voids of accountable size (Figure 1.10b).

3.1.4 Discussion and conclusions

Failure of superheater steel tubes localized either at bent areas of the tubes, or adjacent to the welded joints, exhibiting a characteristic “fish-mouth” or a “cobra” pattern appearance, respectively, advocating the action of creep rupture. The short-time lifetime (~ 15 days of continuous operation) indicated the absence of a primary creep stage, normally occurred in cases of operation under conditions of a high-temperature and/or high stress regime.

- Preliminary assumptions concerning pressure conditions and initial tube dimensions led to maximum tensile hoop stress ($\sigma_{t,\max}$) calculation, estimated at ~ 52 MPa level, exerted at inner surface of the tube, using the Lamé and Clapeyron equation for thick-walled cylinders [30]

$$\sigma_{t,\max} = p_i \left[\frac{r_0^2 + r_i^2}{r_0^2 - r_i^2} \right] \quad (1.13)$$

where r_0 is the outer tube radius (mm), r_i is the inner tube radius (mm), and p_i is the internal pressure (MPa).

- The original design of the pipeline foresees applied thermal loading of 35-40% of the absolute melting point; while during service this parameter reached almost 45% of the absolute melting point of the metal.

- In both failure cases, rupture always occurred within the low-alloy ferritic steel area, probably corresponding to 13Mo3 steel grade; while the presence of various inclusions, such as MnS stringers and round Al_2O_3 , CaO particles, did not seem to have been contributed to the premature failure.

The above evidence suggests potential review and redesign of the continuous operation of pipeline taking into consideration the heat transfer requirements and calorific value of lignite raw materials for the combustion chamber of the unit. The 13Mo3 low carbon steel grade should be likely revised and substituted by more heat-resistant steel grades as superheater construction materials. Regarding the bent tube failures where cracks/separations were observed at the inner bent area, the influence of geometry change on creep life for pressurized pipe bent areas has been extensively reported [22], mainly using the numerical approach by FEA.

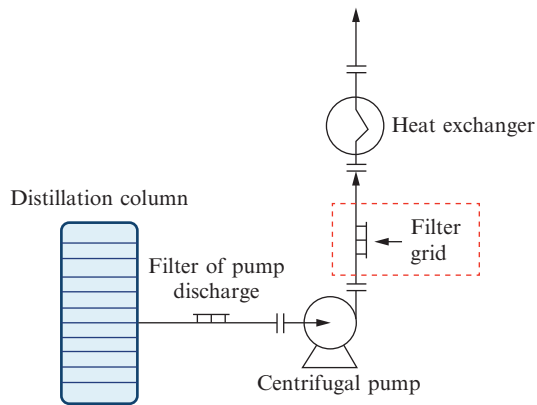
3.2 CORROSION FAILURE OF A STAINLESS STEEL FILTER GRID OPERATED IN A PETROCHEMICAL PLANT

3.2.1 Summary

The present case study concerns the failure analysis of an austenitic stainless steel grid, operating in a shell-and-tube heat exchanger of a petrochemical industry [24]. Macroscopic examination of the grid indicated extensive brittleness and severe cracking perpendicularly to applied stress, while both grid surfaces as well as the interior of the filtration holes were covered significantly by scale deposits. Metallurgical investigation of selected grid areas and surface deposits examination indicated severe cracking exhibiting multiple branching, which suggested strongly the action of SCC mechanism. Besides the extensive cracking areas, pits surrounded by twinning and slip bands were observed, while elemental microanalysis carried out in the areas around and in the interior of the pits indicated mainly the presence of chromium-rich oxides. The catastrophic failure of the stainless grid occurred as a synergistic effect of acting degradation mechanisms, such as SCC, high-temperature oxidation, while the potential formation of detrimental σ -phase cannot be discounted.

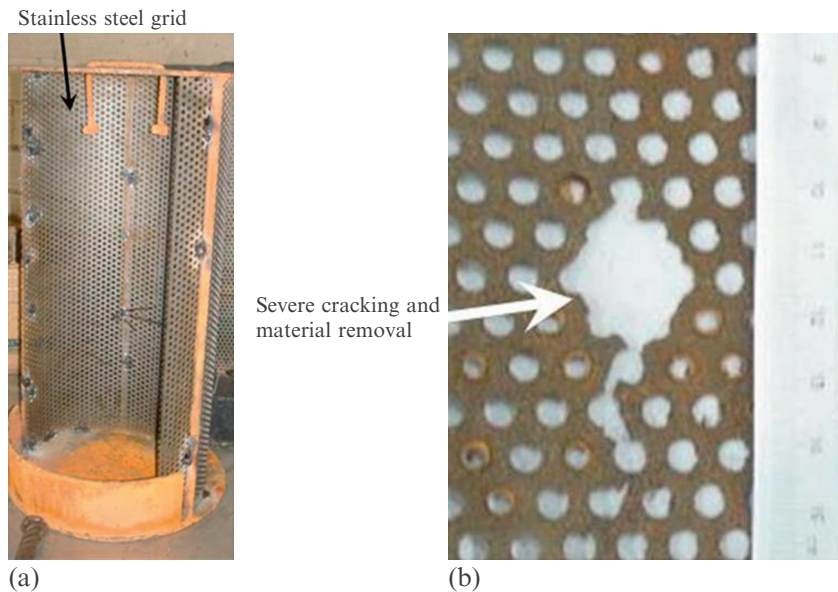
3.2.2 Background information

The products coming from crude oil refinery reach constant temperature conditions for their further postprocessing, commonly achieved using appropriate heat exchange units. The present shell-and-tube heat exchanger, used for the cooling of a naphtha stream with the simultaneous heating of a crude oil, is illustrated in Figure 1.11. Naphtha flows in the tubes (inlet temperature: 137 °C, outlet temperature: 90 °C) of the heat exchanger and crude oil in the shell (inlet temperature: 15 °C, outlet temperature: 74 °C). The operating pressure of the shell was up to 20 bar, while the corresponding pressure of the tubes was up to 7 bar. The present investigation aims to outline the degradation phenomena contributed to the failure of a perforated austenitic stainless steel screen (grid) of a T-type pump (Figure 1.12a).

**FIGURE 1.11**

Schematic of process flowchart indicating the location of the failed filter grid.

(See also Ref. [24]).

**FIGURE 1.12**

(a) Image of standard austenitic stainless steel filter grid before its installation. (b) Macrograph showing detail of the failed filter grid after short service duration (see also Ref. [24]).

The frequent formation of naphtha originated scale deposits inside the tubes which had a significant influence of pressure drop, led to the decision of the installation of a filter between the outlet of the feeding pump and the inlet of the heat exchanger. After its installation, frequent interruptions of the unit (almost once

per 3 months) for filter replacements were recorded due to the failure of the stainless steel grid. It should be mentioned that similar failure incidents of steel strainer used in the filtration of liquid naphtha streams in the petrochemical industry were presented in Ref. [31].

3.2.3 Failure analysis

Elemental microanalysis of the stainless steel performed by EDS, concerning mainly the principal alloying elements (Cr, Ni), showed that the construction material matches approximately the AISI 304 stainless steel grade (see Table 1.4). Macroscopic observation of the failed grid after its removal from the operation site showed clearly pronounced fragility and facile metal removal due to severe degradation processes involved (Figure 1.12b).

The cracks were in a direction normal to the perpendicular axis of the grid which coincides with the direction of mechanical loading due to the component's own weight and the fluid's pressure. In addition, severe fouling and contamination covering almost all the flat surface as well as the interior of the holes was observed. Preliminary observations indicated the nonmetallic nature of the contamination scale, which likely have formed as the result of the interaction between the flowing medium and the grid material.

3.2.3.1 Stress corrosion cracking

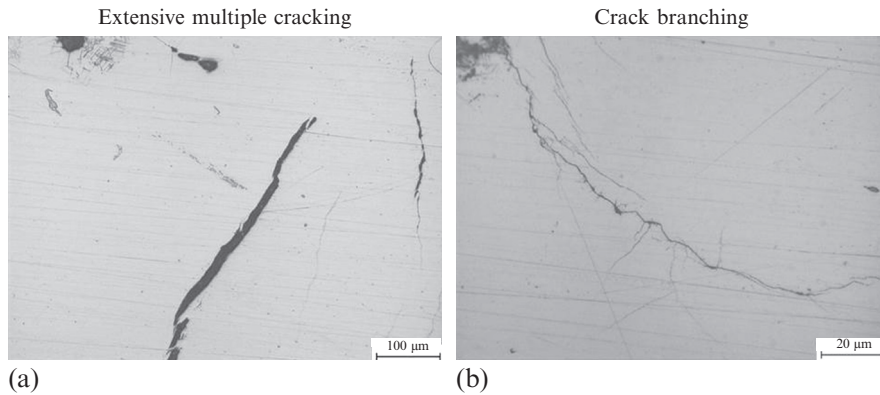
Figure 1.13 shows optical micrographs from representative cross-section areas of the holes of the failed grid. Extensive microcracking was evident exhibiting multiple branching, a typical fingerprint of SCC. Cracks emanated from the circumference of the holes and propagated in the interior forming also secondary cracks. Crack propagation most likely follows a transgranular direction as it is commonly met in austenitic stainless steels.

As a first stage, a crack was initiated at pitted areas close to the surface area where the metal had been exposed to the liquid naphtha environment (at 130 °C approximately), see also Figure 1.14. Pits constitute significant stress concentration areas and thus most sensitive to the nucleation of fatigue or stress corrosion cracks. In the second stage, the crack propagated through the material forming branches, consistently to SCC mechanism. In the final stage, crack propagation led to ultimate catastrophic fracture of the grid.

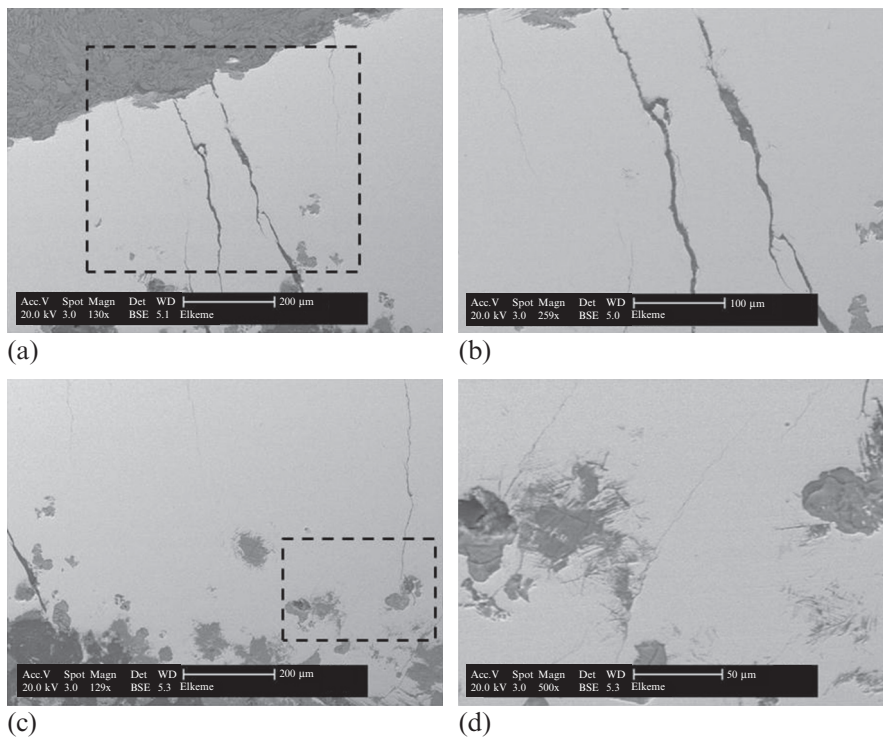
Table 1.4 Chemical composition of the grid steel obtained by EDS analysis (wt%) [24]

Material/Standard	C	Cr	Ni	Mn	Si	Fe
Specimen	–	18.2	8.2	–	–	Balance
AISI 304 grade (UNS S 30400) ^a	≤0.08	18.0-20.0	8.0-10.5	≤2.00	≤0.75	Balance

^aStahlschlüssel (Key to Steel), Verlag Stahlschlüssel Wegst, GmbH, D-71672 Marbach, 2004.


FIGURE 1.13

Optical micrographs of a cross-section area of the failed grid, showing (a) extensive and multiple cracking and (b) crack branching.


FIGURE 1.14

Optical micrographs on cross-sections, demonstrating the initiation and propagation of cracking at areas close to the grid hole surface, resulting in finally catastrophic fracture of the entire grid component. (b), (d) Details of the framed areas presented in (a) and (c) respectively.

Elemental microanalysis showed the presence of mixed Fe/Cr oxides and scale products in the interior of the cracks, suggestive for the corrosive action of the flowing medium.

Deterioration of the material has been associated also with the formation of pits (voids) of average diameter 30 μm (Figure 1.15). As it is mentioned above, cracks emanated from pits and further crack growth and “linking” of individual pits lead to severe metal detachment and final failure. Microstructure features at the vicinity of these voids (as-polished condition) are consistent to twinning and slip band formation (Figure 1.15). These characteristics could be attributed to stress fields generated due to the thermomechanical loading deformation imposed by the operation. Slip bands signify the presence of strain fields which constitute also another indicator of SCC likelihood. In relevant studies, similar “slip band”-type formations are correlated with the precipitation of Cr-carbides. In case of AISI 310 stainless steel, acicular chromium carbide precipitations were reported [25], while in case of AISI 304 stainless steel, the presence of Cr-carbides favored at slip bands and twin boundaries [9]. This effect could be considered as a secondary degradation mechanism since Cr-impoverishment renders the matrix vulnerable and less corrosion resistant. In addition, strain-induced austenite destabilization and martensite transformation at the

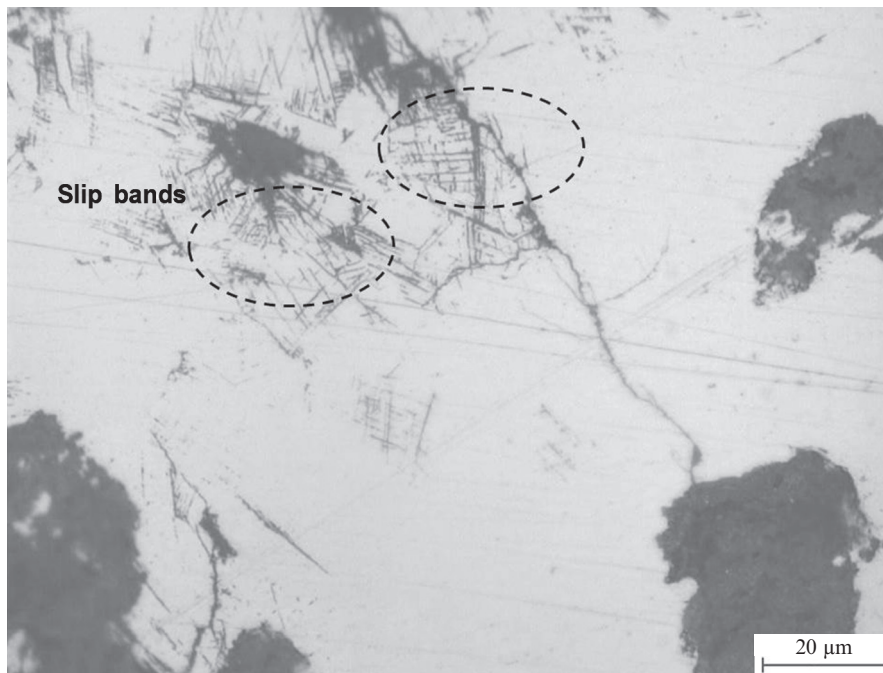


FIGURE 1.15

Optical micrograph showing the presence of pronounced slip bands close to the location of corrosion pits.

slip band areas cannot be ruled out, as strain-induced martensite formation is a common phenomenon in austenitic stainless steels. This can effectively aggravate cracking conditions and facilitate crack propagation, due to the inherent brittleness of the martensite phase.

Thermal degradation phenomena also involve the formation of σ -phase. This is normally related, in austenitic stainless steels, with the tendency of Fe to react with Cr forming an intermetallic compound, during exposure at temperatures higher than 600 °C, resulting in severe loss of ductility and fracture toughness of the steel [26]. The coexistence of σ -phase and Cr-carbides, which result in embrittlement effects on the base metal, has been predicted thermodynamically, during aging of an AISI 310 stainless steel from 600 to 900 °C temperature range; the relevant stoichiometry of σ -phase and Cr-carbide were found dependent on the aging temperature and holding time [27]. A transformation mechanism $\gamma \rightarrow M_{23}C_6 \rightarrow \sigma$ was also referred to outline the intergranular precipitation of σ -phase in austenitic steel [32].

Presumably, the maximum normal working temperature (around 140 °C) does not suggest favorable conditions for the formation of neither σ -phase nor Cr-carbides.

3.2.3.2 Selective/preferential oxidation

Damage grid surface revealed localized scale deposits on the pitted areas with various morphologies associated to nonmetallic reaction products (Figure 1.16). EDS microanalysis identified areas rich in chromium oxides, which also exhibited a brittle behavior (Figures 1.16a and 1.17a), and areas containing complex reaction products originated from the flowing medium (Figures 1.16b and 1.17b). The presence of Cl constitutes an additional supportive evidence of the occurrence of SCC mechanism as a failure contributor.

Selective oxidation leading to the formation of chromium oxides was observed mainly around the middle area of each separate grain, attributed to the nucleation of chromium oxide favored by Cr migration from the interior of the grain toward the

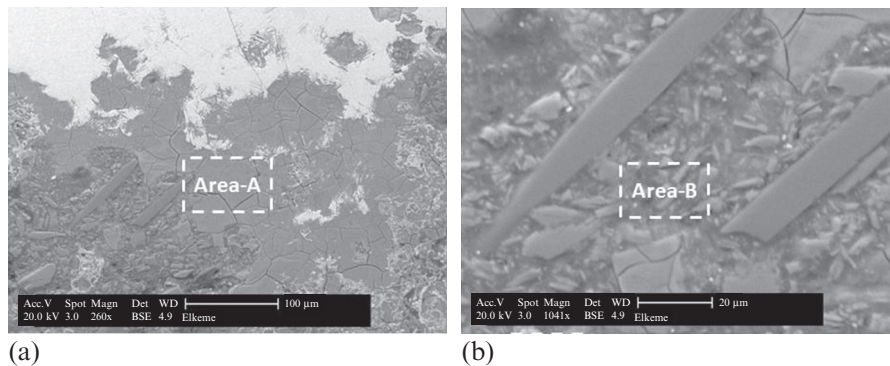
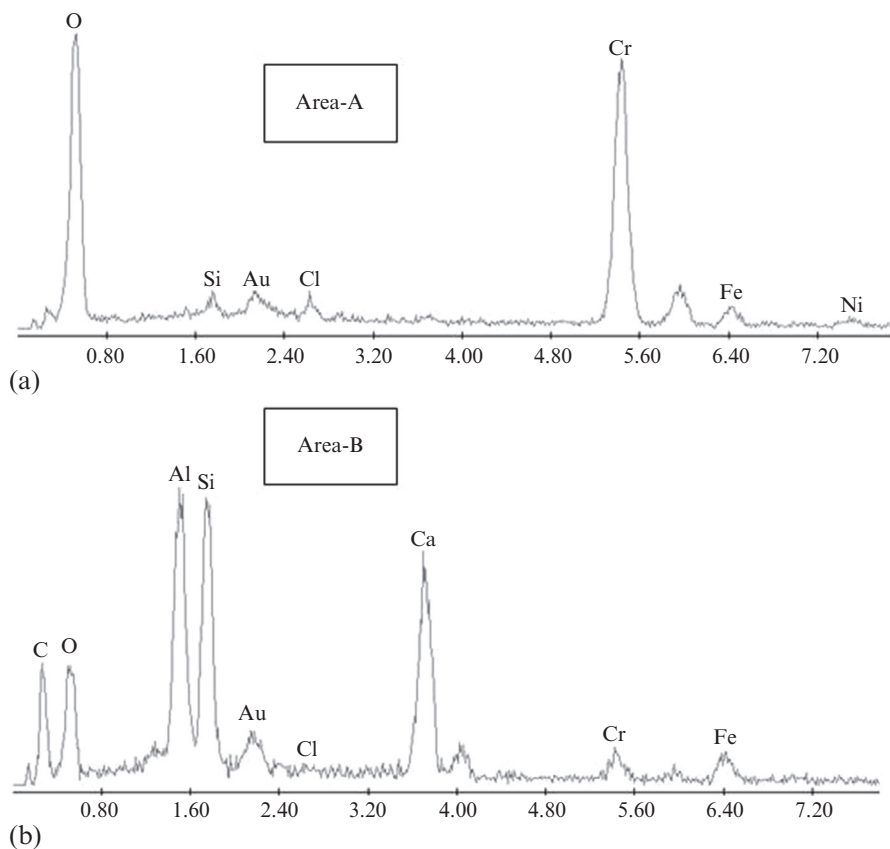


FIGURE 1.16

(a) SEM micrograph showing the morphology of typical corrosion and scale product deposits in the interior of the pits; (b) detail of (a).

**FIGURE 1.17**

EDS elemental microanalysis spectra obtained on selected areas within the filled pits (a) Area-A and (b) Area-B (see [Figure 1.16](#)).

exposed surface. Growth of chromium oxide layer up to a critical thickness leads eventually to crack initiation, due to the imposed tensile stresses, which may result in flaking and removal of oxide layers and exposing the fresh metal surface continuing the oxidation cycle.

3.2.4 Discussion and conclusion

Failure investigation of an austenitic stainless steel grid, operating in a shell-and-tube heat exchanger of a petrochemical industry, was a result of a series of synergistic degradation mechanisms attributed to oxidation and SCC. These mechanisms assisted to severe surface pitting and multiple transgranular cracking, contributing to adverse effects on the structural integrity of the grid and leading to excessive deterioration, fragility/brittleness and subsequently gross metal loss.

The obtained evidence suggested the operation of SCC as the dominant damage mechanism. The SCC failure mechanism is also facilitated by other contributing phenomena, such as selective/preferential oxidation and Cr-depletion. The synergy of aggressive environment, containing corrosive species (such as Cl) at high temperatures, provides an additional driving force supporting the hypothesis of SCC as the principal failure mechanism. Transgranular cracking with multiple secondary branching constitutes typical fingerprints of SCC fracture mode in austenitic stainless steels (see also [Figures 1.13](#) and [1.14](#)). Revision of the operating conditions that foresee pressure attenuation in the inlet fluid stream could be recommended as further corrective action, reducing the filter operating load and thus the risk of SCC. The installation of a baffle system or a flow dividing system that could balance the flow more uniformly could potentially lead to uniform pressure distribution applied on the grid area.

The replacement of the currently used austenitic stainless steel with another alloy of higher resistance in SCC and enhanced high-temperature microstructure stability, such as nickel superalloys (Monel or Hastelloy), could be recommended. Additionally, a ferritic stainless steel, such as the AISI 403 (UNS S 40500), with Cr level within the range 11.5–14.5 wt%, could also constitute an alternative construction material of selected components in chemical process industry.

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Engineering failure analysis in chemical process industries

2

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[‡]The author has been died on January 03, 2015.

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1 INTRODUCTION

The engineering materials are utilized into useful products which are molded and designed to serve mankind for some specific purpose to fulfill needs. The job starts from design engineer who conceives idea, makes plan, and manufactures product to meet intended requirements for some specific service life. In other words, efficient performance of a product depends on design, manufacturing, operation under predefined conditions, and periodic maintenance according to specifications. All these aspects determine the reliability and durability of product under set criteria. Any fault, shortcoming, or flaw at any stage could lead to failure. It may be precluded that the failure of a component at any stage of its lifecycle could happen which may result in decrease of process efficiency, unintended shutdowns, time wastage, loss of money, and in most sever scenario loss of human lives [1,2]. The deviation from the conceptual design, prescribed manufacturing steps, and non-recommended operational conditions could lead to failure and the process is illustrated in Figure 2.1.

From the above discussion, it can be deduced broadly that failure could happen due to following reasons:

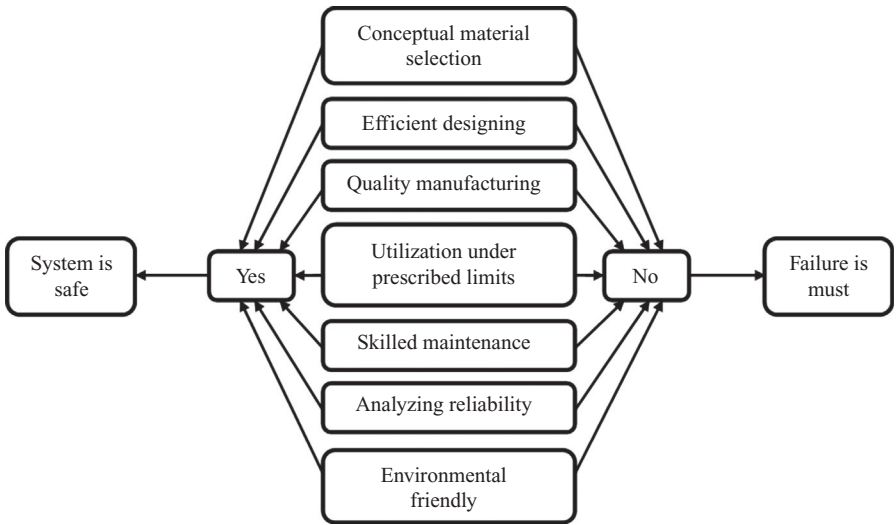


FIGURE 2.1
Overview of general requirements for safe system operation.

1. Wrong material selection
2. Shortcomings in product designing approach
3. Non-engineered manufacturing
4. Unexpected environmental conditions
5. Unscheduled and nontechnical maintenance approaches

The optimum materials selection for a particular component, the design necessitates specific characteristics of materials and prescribes a set of conditions for safe operation. The inadequate mechanical properties, microstructural characteristics, nonuniform alloying elements distribution, and imperfections in the materials are detrimental to the performance of a product or an equipment. Other factors i.e. cost, availability, and fabricability of a material also play a vital role in the selection process. The incapable materials are selected due to unawareness of materials characteristics, non-certified materials, or compromise of the procurer with the quality of the product over cost. The safety and longevity of any component is assured at the designing step of a component. The intelligent approach and consideration of surrounding effects on the designed component is the key responsibility of the designer. The factor of safety should be limited to restrict overdesigning but should be enough to withstand severe conditions throughout its designed life. Design-oriented failure problems may also arise if anisotropy of materials, ignoring service conditions, that is, temperature, pressure, corrosivity of the environment, fluctuating mechanical stresses, and presence of residual stresses are not considered carefully [3]. The conceptual design is always required for critical consideration of sharp notches, abrupt geometry changes, overlooking the compatibility of thick and thin sections with the operational stresses, and chemical/electrochemical attack to avoid failures. The manufacturing of components and fabrication of equipment is equally important as designing. Any wrong action during fabrication or deviation from the mandatory requirements of the adopted standard could also lead to premature failure. The efficient fabrication methods, proper heat-treatment procedures, precise testing, and expert approach to certify equipment for intended application is also needed at the manufacturing stage. It should be considered carefully that unexpected conditions may arise during service beyond the designed limits, it is therefore proper material selection, the best designing approach, and standardized fabrication procedures are essential. This could be beneficial to enhance the service life of a component and or an equipment. During service of the equipment, the standard operating procedure should be strictly followed. The scheduled shutdowns and proper maintenance is required to ensure the integrity of the component and to assess the remaining life [4]. The negligence at any stage or tolerating even minor requirement could lead to failure in a process industry which imposes huge economic burden additive to replacement cost by loss of product, prolong shutdowns, contamination of the product, and decrease in production rate. Operation beyond the prescribed limits of design to acquire large production in short time could also be a threat to the safety of equipment even if all the steps from designing to manufacturing are carefully considered.

One of the major reasons of metallic components failure would be environmental deterioration during service. This may include stress-induced cracking, selective

dissolution of microstructural phases, biologically influenced corrosion, localized attack by the species in the environment, erosion, gouging, cavitation, chemical attack, etc. In addition to the techniques used for metallurgical failure analysis, the electrochemical testing for predicting behavior of same material in simulated environments could also be beneficial to define mechanism. The accurate prognosis to define failure cause needs an expert and a conceptual approach. An analyst can trace back to the product history by closely looking at one or more factors involved in its lifecycle which could also disclose any likelihood of premature failure.

The findings of failure analysis and knowledge of prior damages can be useful in the designing of reliable products, modifications in the manufacturing processes, and developing safe operating procedures. Furthermore, the analysis reports could be beneficial for a materials engineer to identify deficiencies in the existing structural materials, provide guidance to look at alternate materials, and/or improve the properties. The equipment in a chemical process industry during operation experiences extended thermal and intensive pressure cycles which could lead to deterioration of materials. The complex nature of stresses and adverse environmental factors could induce deleterious corrosion effects on the performance of engineering materials and hinder the continuity of operations.

It has been shown in Figure 2.2 that major contributors to accidents in the past were human errors, process control, manufacturing/installation, materials selection,

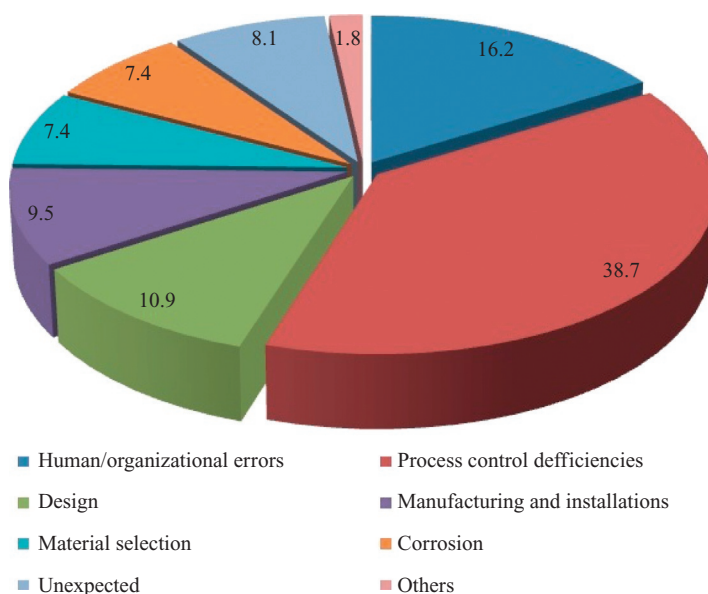


FIGURE 2.2

Major factors that contributed to failures in chemical process industries.

Data extracted from table 2 of Ref. [5].

corrosion, unexpected factors and utilities, etc. [5]. Few case studies describing failures happened in the past that have been listed in Table 2.1 for further reference.

The significant number of failures (38.2%) in the chemical industries was related to the deficiencies in the process control. On the other hand, unforeseen

Table 2.1 Few Failure Case Studies Categorized with Respect to the Possible Cause

Improper material selection and inferior properties
(a) Failure of stainless digester shafts in a paper production plant [6]
(b) Failure analysis of PSV spring in offshore gas production pipeline [7]
(c) Failure of crude oil heater tubes [8]
Process conditions and operational negligence
(a) Failure analysis of stress corrosion cracking in heat-exchanger tubes during start-up operation [9]
(b) Failure in a high-pressure feeding line of an oil refinery due to hydrogen effect [10]
(c) Failure of a super heater tube [11]
(d) Analysis of ethylene cracking furnace tubes [12]
(e) Remaining life assessment of platforming reactor vessels [13]
(f) Failure analysis of heat-exchanger fluid vapor phase vent lines in a polyester plant [14]
(g) Failure investigation of heat-exchanger plates due to pitting corrosion [15]
Manufacturing faults
(a) Weld failures in sleeve reinforcements of pipelines [16]
(b) Fatigue failure of a boiler feed pump rotor shaft [17]
(c) Failure of a valve rod caused by improper heat treatment and metallurgy [18]
(d) Failure analysis of water pump shaft [19]
Deficient process design
(a) Erosion-corrosion of a stainless steel distillation column in food industry [20]
(b) Failure analysis of an electric arc furnace off-gas system [21]
(c) Materials testing of ruptured lead-lined tank and lead pipe [22]
(d) Corrosion failure analysis of spent caustic drainage line pipe [23]
Environmental degradation
(a) Failure of reformer tube of an ammonia plant [24]
(b) Failure analysis of the 304 stainless steel tube in a gas analyzer [25]
(c) Application of gray relational analysis for corrosion failure of oil tubes [26]
(d) Failure analysis and prevention in SCC and corrosion fatigue cases [27]
Maintenance-related issues
(a) Applications of maintenance optimization models: A review and analysis [28]
(b) Risk-based maintenance (RBM): A quantitative approach for maintenance/inspection scheduling and planning [29]
Unexpected factors
(a) Failure analysis and safety evaluation of buried pipeline due to deflection of landslide process [30]

operational conditions, improper material selection, human mistakes, and insufficient knowledge of corrosion threats also account 68-72% of total failures. The periodic inspections, monitoring and maintenance procedures could definitely reduce the risks of premature failure of plant, and hence would insure the safety and reliability of a plant [28–29].

2 HOW TO CONDUCT FAILURE INVESTIGATION

In a broad spectrum, the corrosion failure analysis is defined as investigating the reasons for impairment of a part or a device by the environmental attack. The attack may be more intense due to the existence of inherent abnormalities in the microstructure, irregularities in geometry, processing defects, and any variation in operating conditions.

The incompatible joining of materials, mismatch in material of mating surfaces, overloading, improper repair procedures, and faulty design consideration could also reinforce environmental degradation.

The sequence of failure investigation specially in a chemical processing industry is not similar in every situation and depends on the conditions. Even the resembling historical case studies preclude the identification of factors due to the complicated environmental conditions. The one way of investigation could be adopted as discussed below but not necessarily be true in every case [31].

2.1 UNDERSTANDING THE PROBLEMS

The term “failure” describes the loss of function, service life, and/or fracture which may impair the functionality and limit safe operation in a chemical process industry. The initial statement of a client is useful for an analyst to categorize failure and to investigate failure causes which usually address the nature of problem and could customarily be related to the analysis previously conducted. It is also imperative to gather more and more background information of the process, equipment, and/or system before collecting information and samples from the incidence site. It is also the responsibility of a client to bring analyst in time and should disclose the reality and possible reasons of the failure. The cooperation of a client in this regard could eliminate the risk of misunderstanding and misdirected root cause analysis. It also shortens the time for investigation by reducing extensive efforts which might be unwanted. Hiding the reality could seriously divert the attention of an analyst by missing out the true failure reason and possibly leading to the similar misfortune in future. After collecting detailed information of the accident, the next step for an analyst is to scrutinize design information, service history of the system or component, reviewing maintenance records, monitoring environmental and process transient conditions.

2.2 ONSITE INVESTIGATION

The field visit of a failure analyst at the accident site is always helpful to examine the extent of damage and in collecting evidences. Otherwise the failure study is always dependent on the information provided by the client, operators, and observers. Before conducting survey of accidental site, the analyst should have updated information and have sufficient knowledge about the equipment and other integrated components of the system. For an analyst, the field inspection could be beneficial to narrow down the possibilities of failure causes. Also the documentation of failure site, that is, photography, sketching, and dimensional measurements could be helpful later on during the study. The examination of failure site, “something is considered insignificant sometimes” or even some aspects are overlooked which could become essential afterward. However, adequate photography gives assistance to capture clues and signs of failure stimuli. It is also important to include photographs in the final report of failure as evidence of narrative description. It would be useful to decide collection of important parts or wreckage of the failed components [32]. It provides judgment to the investigator for selecting proper samples which need to be further examined in a laboratory. The careful examination of the failed components, surroundings, and condition of connecting facilities is also essential. All these intelligent efforts made at the accident site are always beneficial to predict the factors involved in failure. To know about any variation in the process parameters or alarming situation before failure, the interview with operators and discussion with eye witnesses could also be helpful to predict probable cause.

2.3 SELECTING THE SAMPLES

For detailed study of the failed component, if possible, the whole system should be transported to the laboratory. Otherwise, the failed part should be separated by cutting sufficiently away from the sound region. If the cracks are present at numerous sites, the samples should be selected which apparently had experienced severe but different operating conditions. If the failure happens by fracture, the mating surfaces should be cut well away to avoid any thermal or mechanical damage of the fractured surface which may obscure the failure evidence [32,33]. The failed samples should be carefully stored in air-tight bags to prevent contamination or oxidation by the environment during transportation to the laboratory.

2.4 EXECUTION OF LABORATORY TESTING

Laboratory testing is an important step in the process of failure analysis. The selection of test methods is very much critical as the results are depictive of materials properties. It is well known that man, machine, and method combination should be experienced, calibrated, and standardized, respectively.

In corrosion failure analysis, the laboratory testing depends on the type of environmental conditions and surface features of failed component. In laboratory, the

protocols for analyzing failed components are included but not limited to the following steps [34–36]:

- Review of background information related to manufacturing, design specifications, process conditions, and scheduled maintenance records
- Primary visual or non-destructive testing to evaluate surface or bulk discontinuities
- Chemical analysis of material, corrosion product, and assembled components
- Monitoring of possible environmental factors which could induce deleterious effect on material characteristics, that is, electrochemical, mechanical, and physical, etc.
- Identifying, selecting, and preserving components in order to avoid damage of surface features which could obstruct correct interpretation of failure mechanism
- Conducting standard metallographic examination to determine microstructural inhomogeneities and comparison of sound part microstructures with infected one
- Mechanical testing of failed and exactly same material for comparison
- Stress and design analysis of component on the basis of testing results
- Fractography of fractured surfaces to correlate features with microstructural, stress distribution, and validation of environmental attack
- Electrochemical analysis of failed and unused material in simulated environment to inquire about any deviation

2.5 COMPILING THE DATA

The analysis and compilation of data is as important as the preliminary steps of failure analysis. The sequential arrangement of data is important to understand the cause and mechanism of failure. The presentation of observations and testing results should be simple and elaborative. There should be a connection between each step of analysis and representation of the data. The effective and useful way of reporting the investigation is to convert important data in graphical trends, tables, histograms, and images. The drawings should be constructed to depict the design of component, identifying failure zone, and section of samples analyzed. The report language should be more focused and technical. Keeping in view the importance of failure investigation, the contents should be arranged in such a way that it can be understandable to designer, manufacturer, and user. This document will also be a guide for all of them to select proper material, treatments, and design to avoid such misfortune in future.

2.6 DRAWING CONCLUSION

The collection of relevant background data, field inspection, and conduction of laboratory testing are the critical steps for evaluating failure cause. A thorough background knowledge of equipment or device failed, connecting parts, effect of operating conditions, material behavior, previous history, maintenance or repair

records, apparent surface features, and laboratory testing results are required. In addition, a careful review of each aspect of investigation procedure is important because there might be legal claims involved. The conclusion could also impose serious concerns to designers, manufacturers, and operators. It is therefore a great responsibility over the shoulders of failure analyst to comment on critical issues and highlight major discrepancies in the product or device that caused failure. The conclusion should be concise, elaborative, and effective stating the facts and do not include misleading comments.

2.7 SUGGESTING REMEDIES

Once the conclusion is drawn on the basis of findings and laboratory test results, an expert approach is required to suggest actions to be taken to prevent such failure. It may describe the procedure to either decrease the frequency of failure or to prevent it. The best engineering practice demands the detailed discussion on the all possibilities which could arise in product designing, manufacturing, properties, performance, and operating conditions. The preventive measures and standard procedures should be simple, easy to adopt, economically and technically feasible.

3 CASE STUDIES

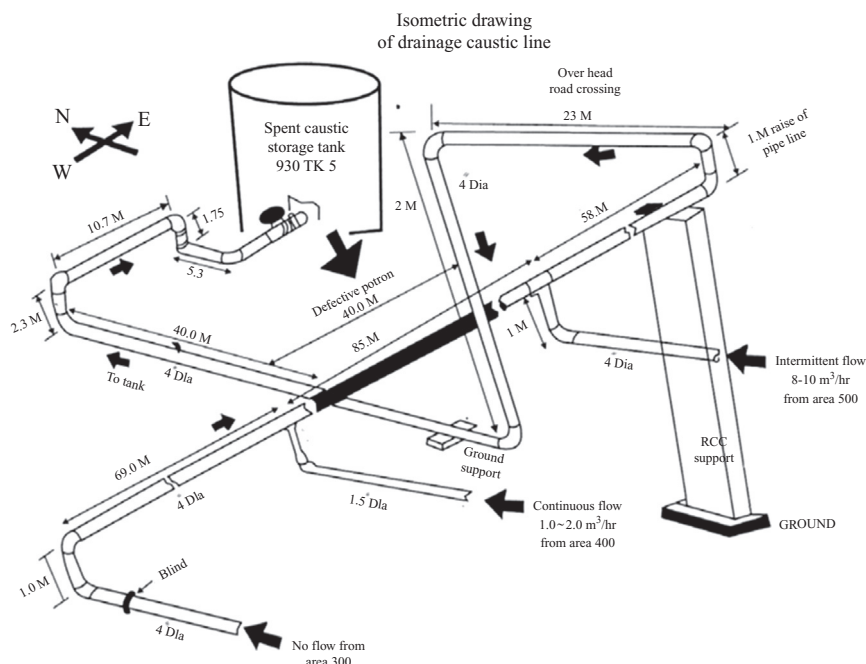
The two failure case studies have been presented here for guidance and as examples. The first study describes the failure of spent caustic drainage line pipe which leaked just after 8 months from installation. The failure was occurring frequently at the same location even after replacement of the line pipe between continuous flow line (CFL) and intermittent flow line (IFL) connected with the caustic storage tank. The second case belongs to the failure of plate heat exchanger integrated with the power generation system of a textile mill. The frequent perforation and fouling of 316L stainless steel plates at open-circuit cooling water-side of heat exchanger were hindering the smooth operation of mill power unit. The brief history, description of investigation procedure, causes, and remedial suggestions are provided in the following sections.

3.1 STUDY OF A SPENT CAUSTIC DRAINAGE LINE PIPE FAILURE IN AN OIL REFINERY

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3.1.1 Background

This case study will describe the failure of a 4-in. line pipe connected with a spent caustic storage tank in a crude oil refinery. About 40 m line leaked at many locations due to pitting and this happened in just 8 months after its installation. The spent caustic was designed to flow through this horizontal line by three inlets designated as 300, 400, and 500 as shown in Figure 2.3. The damaged line portion is highlighted in black

**FIGURE 2.3**

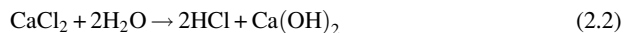
Schematic diagram of drainage caustic line pipe system.

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and the leakage was frequent as exposed by the client and repeatedly happened between continuous (400) and intermittent flow (500) lines.

3.1.2 Brief description of process and experimentation

Crude oils from the production well may contain salts mixture of NaCl, MgCl₂, and CaCl₂. Chemically, NaCl is fully ionized in water and does not hydrolyze or precipitate out significantly. However, the MgCl₂ and CaCl₂ may be hydrolyzed to form HCl as given in the following equations.



The hydrolyzed magnesium and calcium hydroxides are precipitated out and leave HCl in the aqueous phase of crude oil. This is then neutralized by NaOH to form soluble NaCl and is eliminated from the crude oil before preheating. The line pipe system designation and process parameters are given in Tables 2.2 and 2.3, respectively.

Table 2.2 Spent Caustic Drainage Line Pipe System Designations

Designation	Flow Conditions	Pipe Dimensions (in.)	Flow Rate (m ³ /h)
NFL-300	No flow	4	—
CFL-400	Continuous	1.5	1.0-2.0
IFL-500	Intermittent	4	8.0-10.0

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Table 2.3 Process Parameters and Materials Specifications

Parameter	Value
Line service	Spent caustic
Total alkalinity	0.90%
pH	8.3
Temperature	38 °C
Average flow rate	1.2-2.0 m ³ /h
Pipe	4 in. (ASTM A53 Gr B)

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**FIGURE 2.4**

Through holes depicted due to pitting and deposition of corrosion product are present at the inner surface of line pipe.

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The damaged portion of line pipe was selected and cut for visual examination from inside as shown in [Figure 2.4](#). The metallographic study was carried out to evaluate any variation in microstructure due to any metallurgical variation during manufacturing or operation. To investigate the cause of failure, the line pipe design and flow rate calculations were critically examined. The electrochemical behavior of line pipe material was determined by potentiostatic method to predict its corrosion tendency toward spent caustic solution.

3.1.3 Visual observations and testing

1. Visual observation confirmed the formation of holes of varying sizes (5–25 mm) in the line pipe along 40 m length between CFL400 and IFL500 connected flow lines (Figure 2.5a).
2. The pits were observed underneath the layered corrosion product which was reddish brown from the top and grayish black at the bottom.
3. Most of the pitting and holes were formed at the 6 O'clock position and the upper part of the line pipe was unaffected.
4. The microstructural analysis showed the banded structure of fine pearlier stringers along the grain boundaries of ferritic matrix (Figure 2.5b). The hardness of the used material was 44.5HRA. The cross-sectional metallurgical structure showed decarburized ferrite layer at the inner side of line pipe surface as shown in Figure 2.5c.
5. At the pit site, the elongated hair-like corrosion features in the ferrite along the pearlite stringers (Figure 2.5) were observed by metallography, indicating the subsurface enlargement of pits.

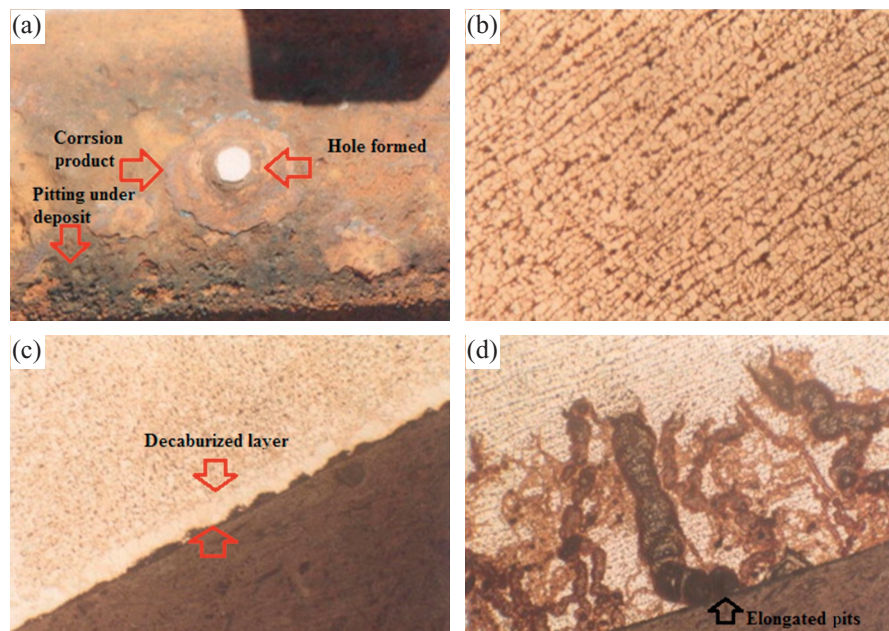


FIGURE 2.5

Visual and microstructural features of failed line pipe. (a) perforation and corrosion, (b) optical microstructure 200 $\times$, (c) decarburization layer, and (d) elongated pits formation.

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3.1.4 Critical review of process design

1. It was determined that no fluid was flowing in NFL300, acting as blind section and there was continuous flow of spent caustic with a flow rate of (V) 1.0-2.0 m³/h from (1.5 in. diameter) CFL400 line. Based on steady state flow conditions, the flow velocity ($v = V/A$) of spent caustic solution was 0.24-0.490 m/s. The flow velocity was reduced to 0.0345-0.0690 m/s when spent caustic enters the 4-in. diameter horizontal line.
2. The intermittent supply through 4-in. diameter IFL500 with a volumetric flow rate of 8-10 m³/h obtained a velocity of 0.276-0.345 m/s at the entrance of damaged line pipe.
3. The intermittent flow of spent caustic when mixed with continuous supply, the resultant flow rate was increased to 9-12 m³/h beyond the “T” junction from IFL toward the spent caustic storage tank and attained a flow velocity of 0.311-0.414 m/s.
4. The reduced flow velocity of (spent caustic from 0.24-0.490 to 0.0345-0.0690 m/s) from the T junction of CFL400 into the horizontal line pipe could facilitate the settling of suspended solids in the solution at 6 O'clock position and concentration aeration cell could form underneath it. Also the reduced velocity at the junction of CFL400 could shield the surface by producing stagnant conditions at the bottom surface of line pipe.
5. The high level of dissolved sodium chloride in spent caustic solution and formation of aeration cells due to deposition of suspended solids accelerated the pitting corrosion. The non-feeding line pipe (NFL300) also promoted the deposition due to reduced flow velocity between two junctions.

3.1.5 Analysis of solutions

The tap water and spent caustic solution analysis results are given in [Table 2.4](#). The spent caustic solution contained higher chloride ions than water.

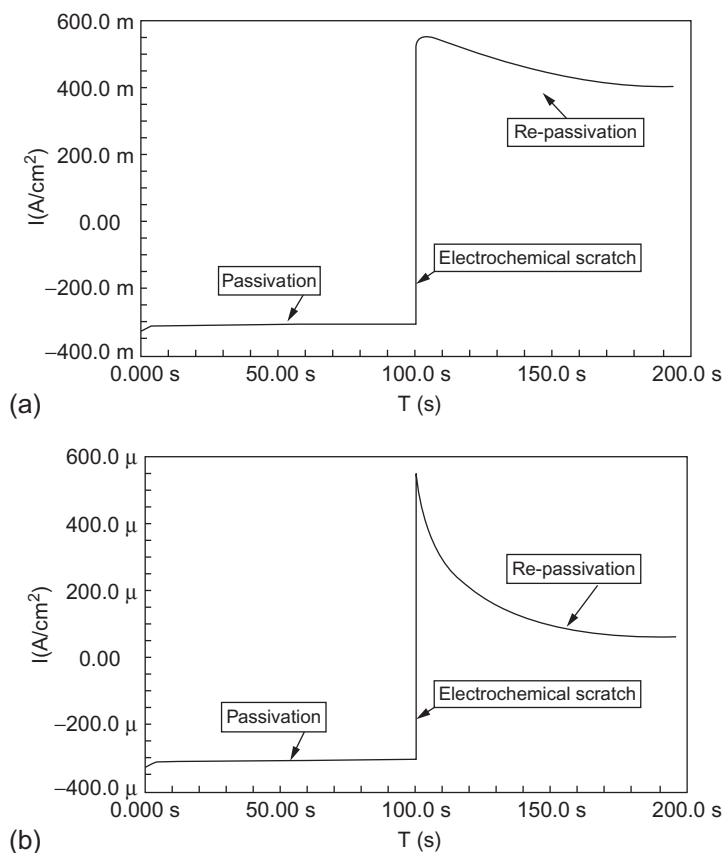
3.1.6 Electrochemical testing

The potentiostatic polarization scan was obtained in spent caustic solution and also in tap water for comparison. It is depicted in [Figure 2.6](#), the tendency to repassivate by the line pipe material in water was higher than in spent caustic. This was caused due

Table 2.4 Chemical Analysis of Tap Water and Spent Caustic Solution

Species	Tap Water	Spent Caustic
Total dissolved solids (ppm)	269	6340
Total alkalinity (ppm)	220	4520
Chloride (ppm)	15	1200
pH	7.1	8.86
Conductivity (μMhos)	382	10,220

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**FIGURE 2.6**

Potentiostatic polarization scans of line pipe material in (a) spent caustic solution (b) tap water.

to the presence of high chloride and alkalinity in spent caustic solution. The electrochemical testing results also validated the possibility of spent caustic attack to the surface and were in support to predict corrosion phenomenon. The passivation was prevented by high concentration of chloride and high alkalinity.

3.1.7 Conclusions

1. The near-stagnant conditions in a horizontal 4-in. line pipe at the junction of CFL400 were formed due to reduced velocity 0.0345-0.0690 m/s, hence settling of suspended solids and corrosion product at 6 O'clock position resulted in the formation of aeration cells.
2. The differential aeration cells underneath the deposit resulted in the breakdown of passivity and presence of high chloride content may produce acidic environment locally at the pit sites.

3. During potentiostatic scan, the line pipe specimen did not attain passivity due to high chloride level in spent caustic solution compared to water.
4. The rate of penetration due to pitting was increased with time as perforation.

3.1.8 Recommendations

On the basis of visual, metallurgical, design consideration, and electrochemical testing results it was suggested to replace 1.5-in. CFL400 pipe with 4-in. diameter pipe to eliminate the velocity-reduction effect in the horizontal line pipe which could also prevent the formation of stagnant layer and hence the deposition of suspended solids. In a continuous flow situation, the formation of local cells will be restricted by the elimination of differential aeration cells.

3.2 FAILURE OF PLATE HEAT EXCHANGER DUE TO PITTING

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3.2.1 Background

The heat-exchanger plates failed due to pitting and perforation. This plate-type heat exchanger was the part of a power generation system and it was designed to transport heat of close-circuit jacket, primary cooling system auxiliary water by cooling with open-circuit cooling tower (CT) water (secondary cooling system) to decrease the temperature by 20 °C. Almost 32 plates were perforated by the secondary cooling water in the open-circuit side of CT system.

In the jacket water side, the plates were arranged sequentially in two pairs, each having two plates one for jacket water and the other two plates for CT water; thus, 24 plates were tightened in a bundle. There was similar arrangement in the auxiliary water side of heat exchanger. The hot and cooling water flow arrangement is shown schematically in Figure 2.7.

Internal lube oil and cylinder head jackets of the power engine were cooled by auxiliary water and jacket water heat exchanger in a close circuit, respectively.

For efficient release of heat from the four power engines and maintaining the appropriate temperature, each was connected with an individual CT having capacity 50 tons/h (Table 2.5).

3.2.2 Analytical Approach

To investigate the reason of perforation and pitting of plate heat exchanger, the following procedure was adopted:

1. Collection of historical, design, and operational data
2. Visual inspection
3. Metallurgical analysis
4. Chemical analysis of every water system (raw water, reverse osmosis (R/O) water, CT feed, and CT bleed water)
5. Investigating the electrochemical behavior of heat exchanger plates in water systems

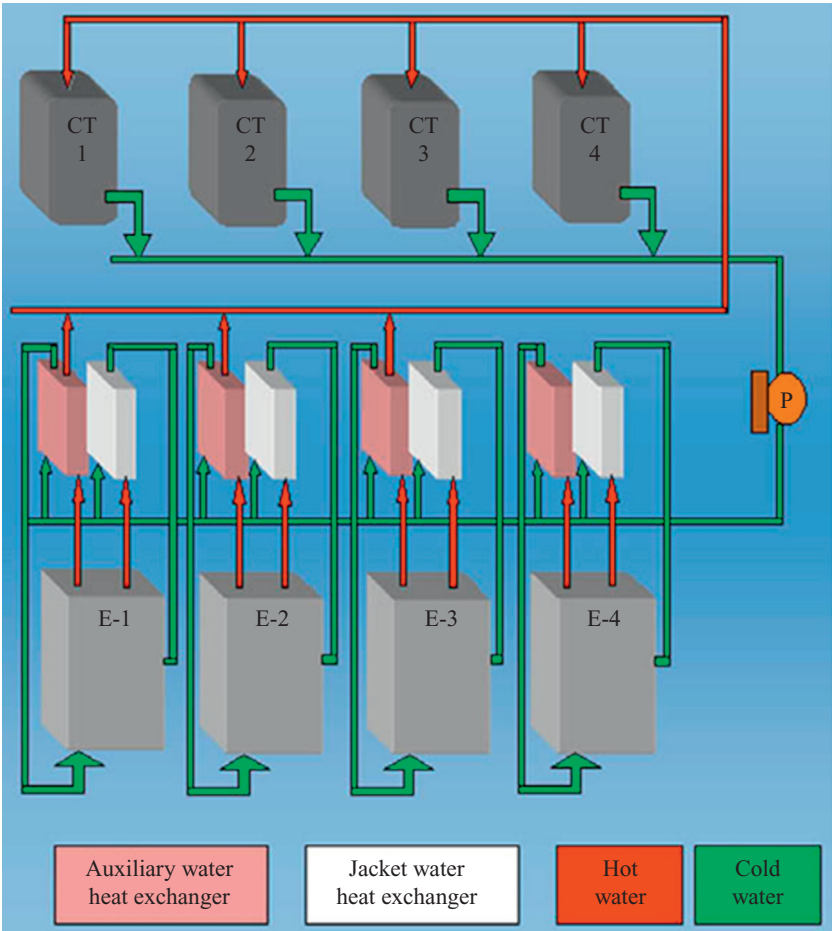


FIGURE 2.7
Flow sheet of a power generation and cooling system.

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Table 2.5 Cooling System Temperature Requirements at Inlet and Outlet

	Temperature (°C)	
Primary Cooling System Water		
	Inlet	Outlet
Jacket water	60	80
Auxiliary water	40	58

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3.2.3 Description of Process

The detail of any chemical addition in CT water and information regarding operational parameters was scrutinized as given in [Tables 2.6](#) and [2.7](#), respectively. The proprietary chemicals were being added in specific amounts into CT water to avoid corrosion, scaling, and algae formation as suggested by the chemical supplier. The pH of the water was being adjusted by adding sulfuric acid in the CT water. It was mentioned in the operating procedure documents that CT water should be free from chlorides, hypo-chlorides, and dissolve solids in excess.

The CT process water specifications recommended by the designer are given in [Table 2.8](#). In close-circuit cooling system, the proprietary chemical (NALCOOL 2000) was being used.

3.2.4 Review of the process

It was reported by the client that during the scheduled maintenance after 2000 h operation of power engines, severe pitting and fouling was observed in the open-circuit system toward the CT water side as shown in [Figure 2.8](#). Also, there was severe pitting and blistering in the pipe line of CT. But there was no pitting and deposition of

Table 2.6 Cooling Tower Water Proprietary Additives

Additives	Dose	Concentration
AC 220	(kg/CT/24 h)	0.450
AC 204	(kg/CT/24 h)	0.200
AC 460	(kg/CT/168 h)	2.0
Sodium hypochlorite	(kg/CT/48 h)	0.5
Sulfuric acid	(kg/CT/24 h)	10-15

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Table 2.7 Chemical Analysis of Cooling Water After 3 and 6 Months and Comparison with the Design Limits

Species/Parameters	Design Limits	(3 Months)	(6 Months)
Ca hardness (ppm)	<350	320	320
Mg hardness (ppm)	–	–	–
T. Alkalinity (ppm)	<500	490	460
Chloride (ppm)	–	–	–
Phosphate (ppm)	3.0-6.0	5.0	5.0
Iron (ppm)	<0.5	Traces	0.1
Ryzner Index	4.8-6.0	4.46	–
TDS	2500-3200	2280	3100
pH	7.8-8.6	8.94	8.40

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Table 2.8 Cooling Tower Water Design Limits as Suggested by the Designer

Parameters	Value
System volume	0.5 m ³
Recirculation rate	120 m ³ /h
Temp. drop	06 °C
Water	Reverse osmosis
Evaporation	0.457 m ³ /h
Bleed off	0.015 m ³ /h
Make up	0.472 m ³ /h

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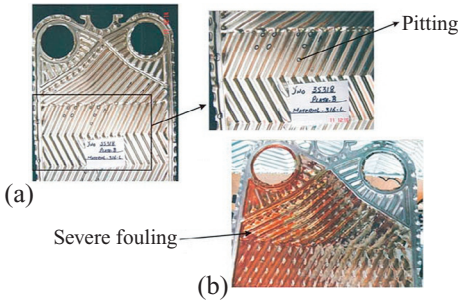


FIGURE 2.8

(a) Pitting and (b) fouling of heat exchanger plate from the CT water side.

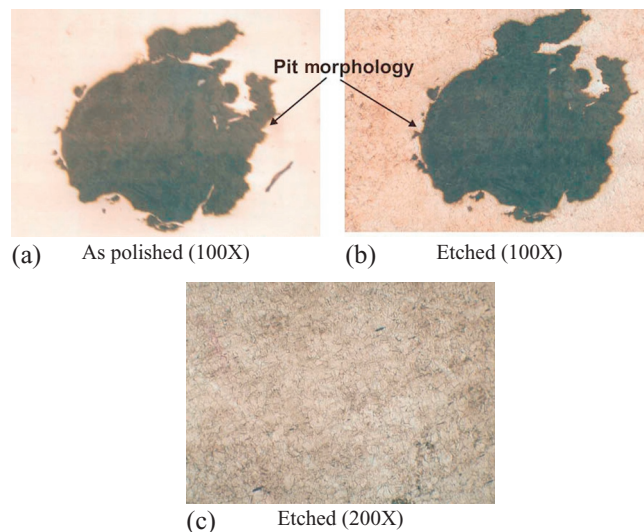
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sludge in the close-circuit side. The perforated plates damaged by pitting corrosion were analyzed metallurgically and electrochemically. Specially, the composition of raw, R/O, CT feed, and bleed water was critically monitored and chloride levels in these water samples measured in the laboratory.

3.2.5 Testing Results and Findings

The analysis of plate samples was confirmed as AISI 316L stainless steel and was in accordance with design specification. There was no heterogeneity in the microstructure and plate material was metallurgically clean having no inclusions and slag particles in it. The microstructural analysis and morphology of pit showed twined and banded structure of austenitic typical to the microstructure of cold rolled stainless steel as shown in Figure 2.9. The pits geometry was irregular and their approximated size was in the range of 0.8-0.95 mm.

The water analysis of CT water confirmed the high level of chloride in CT feed (142 ppm) and bleed water (709 ppm) which was beyond the limits recommended by the supplier and manufacturer of power engines. Also, the client disclosed that the need for addition of treatment chemicals was being determined on the basis of total

**FIGURE 2.9**

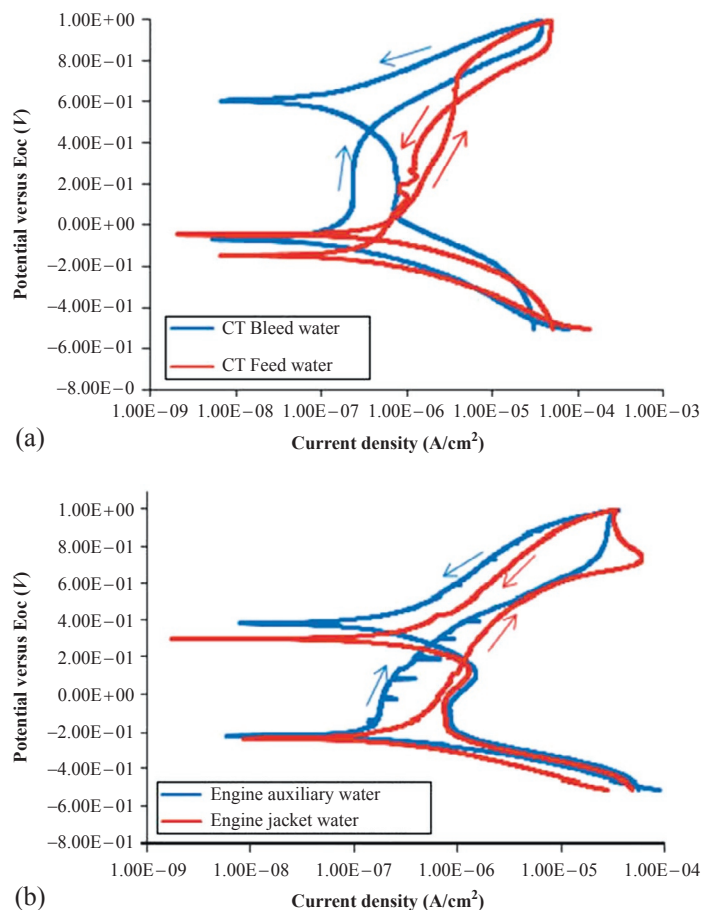
Pit morphology of heat-exchanger plate sample. (a) polished (100 $\times$), (b) etched and microstructured, and (c) etched from sound region (200 $\times$).

dissolved solids (TDS) value than chloride. The level of TDS in raw (2020 ppm) and CT bleed water (3060 ppm) was high which could be the probable cause for severe fouling of the plates and CT piping network.

In laboratory, potentiodynamic cyclic polarization and potentiostatic polarization scans of plate sample (from sound region) were obtained to measure the severity of CT feed and bleed water toward pitting corrosion. The results under simulated conditions are shown in Figures 2.10 and Figures 2.11, respectively. It was measured that corrosion potential of plate sample was more noble (-47.62 mV) than in CT feed water (-71.43 mV). Also, the pitting resistance of plate samples in CT feed water was higher (769.82 mV) than in bleed water (414.56 mV) due to higher concentration of chloride in it. The negative hysteresis in cyclic polarization depicted by stainless steel samples in auxiliary and jacket water corresponded to higher resistance to pitting. The lower repassivation potential in bleed water by potentiodynamic polarization was validated by the electrochemical scratch test (potentiostatic polarization) by plate samples (Figure 2.11). The repassivation current density in CT feed and bleed water was 3.13 $\mu\text{A}/\text{cm}^2$ and 0.28 mA/cm^2 , respectively.

3.2.6 Conclusion

The chemical analysis of metal plates was confirmed as 316L stainless steel and there were no metallurgical inhomogeneities in it. The high level of chloride/chlorine in open-circuit system (CT water) caused severe pitting and perforation in the heat-exchanger plates. The low resistance to pitting by 316L plate samples in CT feed

**FIGURE 2.10**

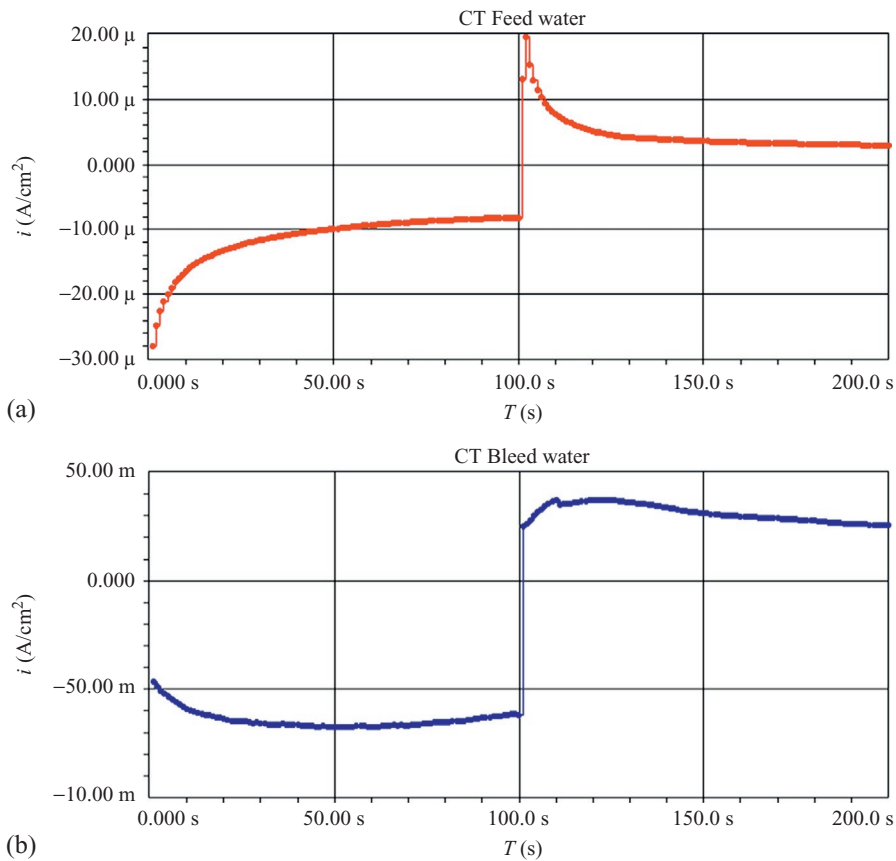
Potentiodynamic cyclic polarization of heat-exchanger plates. (a) CT feed water and (b) CT bleed water.

and bleed water was confirmed by cyclic polarization. The inability of plate samples to passivate in CT bleed water also validated the aggressiveness of environment and hence could be a cause of pitting and perforation of plates.

3.2.7 Recommendations

Keeping in view the process design requirements, visual observations, and electrochemical testing under simulated conditions in laboratory, the following recommendations were advised:

1. The feed water in the CT should be of R/O quality.

**FIGURE 2.11**

Potentiostatic polarization scan of heat-exchanger plates. (a) CT feed water and (b) CT bleed water.

2. The addition of inhibitive (proprietary chemicals) species in the cooling water should be administered on the basis of chloride level rather than TDS values.
3. To avoid localized corrosion and hence perforation, the chloride concentration should be lower than 50 ppm under prescribed operational conditions.
4. Considering operational parameters, chemical dosage requirements, and materials specifications, an expert approach is required for maintaining effective water-treatment process to prevent such failures in practice for such application.

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Boiler tube failures: Some case studies

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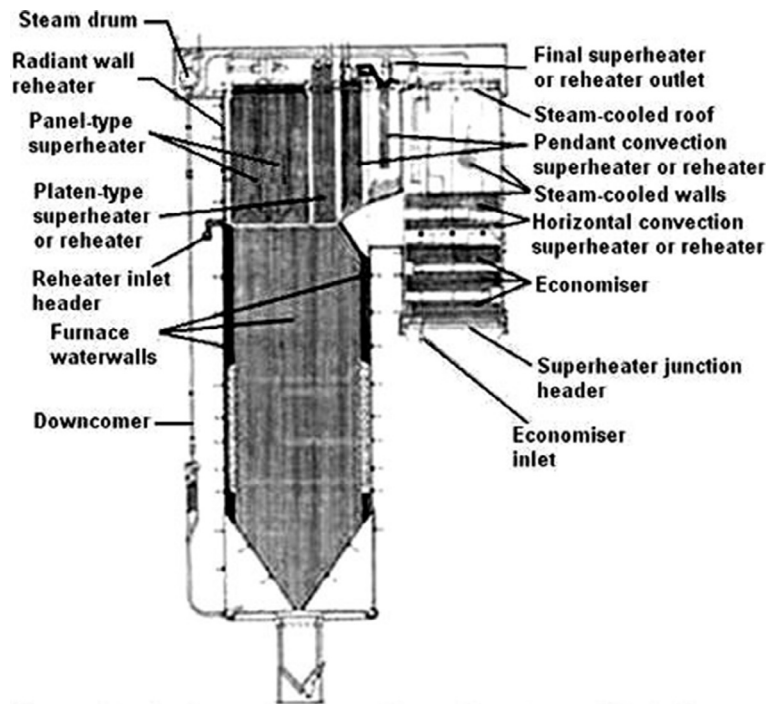
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1 INTRODUCTION

The principal route for producing electricity is the conversion of mechanical energy of rotation into electrical energy using a generator. The rotation of the generator shaft is achieved by coupling it to a turbine where the kinetic energy of the moving fluid is converted into mechanical energy of rotation. The working fluid may be wind, water, steam, or combustion gases. The most common fuel used to produce steam in the boiler is coal. In USA, 70% of the electricity is produced from fossil fuel power plant; in India also, it is about 80%, and the rest from other sources, such as, wind, water, nuclear, etc. [1,2]. The schematic illustration of a boiler is shown in Figure 3.1.

The severe service environment in fossil fired boilers is largely the reason for these failures as the effect of stress, temperature, corrosion, erosion, and vibration combine to produce degradation of the tubing steel. Corrosion by oxidation, by combustion products, and by impure boiler water can significantly reduce the tube wall thickness and result in failure of a tube many years before the expected end of its useful service life.

**FIGURE 3.1**

Typical cross section of large drum-type utility boiler, showing major water and steam-cooled tube circuits [3].

Exposure of the tube steel to elevated temperature causes microstructural changes of the steel constituents [4–6]. As for example, carbon steel will experience spheroidization in which iron carbide particles change from plate or rod shapes to more or less spherical ones. Plain carbon and carbon molybdenum steel will experience graphitization in which Fe_3C particles transform into free iron and graphite molecules. Austenite stainless steel will develop carbide precipitates and intermetallic compounds. Microstructural changes occur even at design conditions with long-term service. Their formations are accelerated when tube steel is operated above design conditions.

Much of work has been carried out and still going on regarding boiler tube failure mechanisms by renowned organizations and individuals. Some of the organizations who are in the field are Electric Power Research Institute (EPRI, USA), Central Electricity Generating Board (CEGB, UK), ASME Boiler and Pressure Vessels, USA, Combustion Engineering, USA, etc. Failure of boiler tubes is the prime reason of boiler outage in most of the countries. Boiler tube failures are of immense concern to utility companies and boiler manufacturers. There are many reports regarding failure modes and mechanisms. Damage mechanisms which mostly contribute the

maximum percentage of tube failures [3,7] are: Short-term overheating, high-temperature creep, caustic corrosion, and hydrogen damage.

1.1 SHORT-TERM OVERHEATING

Stress and temperature influence the useful life of tubing steels operating in a boiler. The strength of a boiler tube is dependent on the level of stress as well as temperature when the tube metal temperatures are in the creep range. Since an increase in either stress or temperature can reduce the time to rupture, attention must be given to both factors when investigation is to be carried out regarding failure caused by stress-rupture mechanism. Stress-rupture failure mechanisms are predominantly experienced in steam cooled superheater and reheater sections where the operating temperature is in the creep range. Stress rupture can also be experienced in water cooled tubing if abnormal heat-transfer conditions exists that results in an increase in the tubes operating temperature [8,9]. If the thickness of the tube decreases by corrosion, or erosion, the hoop stress will increase and hence the likelihood of failure. The circumferential hoop stress in a tube is determined by the diameter and thickness of the tube as stated in the following Equation 3.1.

$$\sigma_H = PD_M/2W \quad (3.1)$$

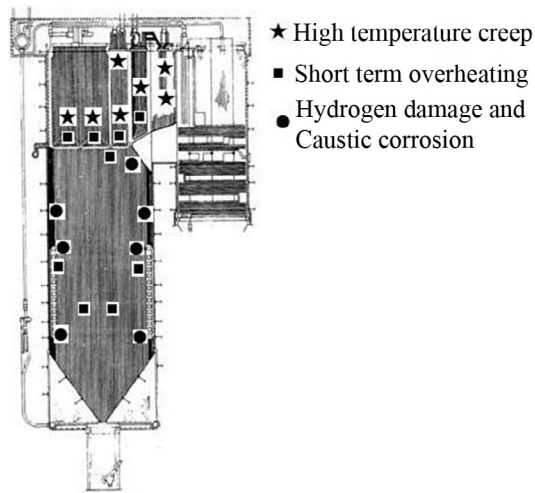
where: σ_H = estimated hoop stress; P = internal stress pressure; D_M = mean tube diameter; W = tube wall thickness.

The “overheating failure” means a failure resulting from operation of a tube at temperature higher than expected in design selection of the tube steel for a period of time sufficient to cause a stress-rupture failure. Time at temperature is an important factor and these types of failures are called “short-term” and “long-term” overheating. Figure 3.2 shows the locations of the boiler where short-term overheating can occur and Figure 3.3 shows thick-edged short-term overheating type failure of a boiler tube.

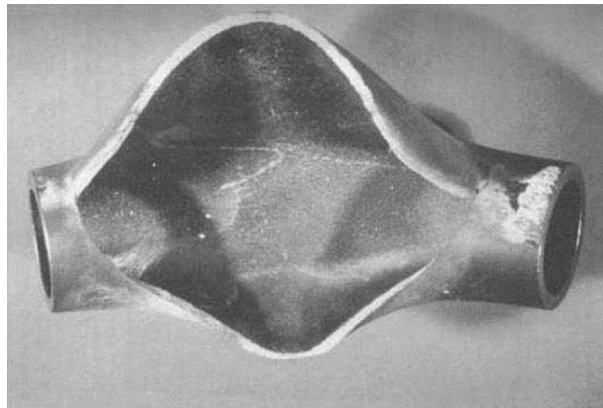
1.2 HIGH-TEMPERATURE CREEP

Boiler tube failure can result from high-temperature creep of superheater and reheater tube steel. Metal degradation and permanent deformation will occur with time depending on the actual stress level and temperature. If temperatures and stresses exceed design selection values, the tube steel will exhibit a higher creep rate and will fail earlier than expected.

High-temperature creep failures are called “long-term” or extended overheating failures. Such a failure results from a relatively continuous extended period of slight overheating, a slowly increasing level of temperature or stress, or accumulation from several periods of excessive overheating. The creep occurs along the grain boundaries of the steel and is aligned 90° from the direction of applied tensile stress. Creep deformation results in little or no reduction in wall thickness but produces measurable creep elongation or increase in diameter in ferritic steel tubes. The creep

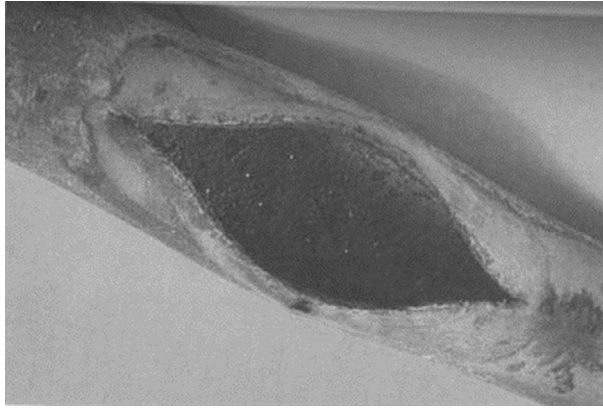
**FIGURE 3.2**

Boiler locations showing different types of failures.

**FIGURE 3.3**

Unusual short-term overheating failure. Excessive swelling but thick-edged fracture surfaces are produced when a tube's metal temperature exceeds its upper critical temperature (Ac_3) and the iron in the steel has been completely transformed from ferrite to austenite [3].

cavitation of the superheater/reheater tube due to exposure at higher temperature and stress was shown by earlier investigators [10–12]. Figure 3.2 shows the locations of the boiler where high-temperature creep can occur and Figure 3.4 shows fish-mouth-type high-temperature creep failure.

**FIGURE 3.4**

Fish-mouth type high temperature creep failure. Typical features of a long-term, general overheating, creep rupture failure include an open-mouthed longitudinal split and thick-edged fracture surface. Oxide scale on the external and internal surfaces also indicated a high temperature creep failure [3].

1.3 HYDROGEN DAMAGE

Boiler tube failure caused by hydrogen damage resulted from liberation of atomic hydrogen during corrosion process [13,14]. This hydrogen is capable of diffusing into the metal wall. It may pass through the wall into the furnace environment. If the atomic hydrogen reacts with carbide (Fe_3C) phase of tube wall metal to form methane (CH_4), $4\text{H} + \text{Fe}_3\text{C} = \text{CH}_4 + 3\text{Fe}$, the outward diffusion ceases and methane begins to accumulate within the tube wall. The accumulation sites for the CH_4 are grain boundaries adjacent to colonies of pearlite, the Fe_3C containing constituent of the microstructure. CH_4 gas pressure at the grain boundaries eventually exceeds the grain boundary strength, resulting in short, discontinuous, randomly oriented microcracks. Interlinkages of these microcracks diminish the load carrying cross-section of the tube wall metal. Eventually, a thick-walled rupture of the tube metal occurs. Figure 3.2 shows the locations of the boiler where hydrogen damage can occur and Figure 3.5 shows hydrogen damage, window opening, and thick-edge-type failure.

1.4 CAUSTIC CORROSION/GOUGING

Caustic corrosion/gouging occurs when alkalinity of boiler water increases. Caustic corrosion is also called caustic attack. Caustic corrosion develops from deposition of feed water corrosion products in which NaOH can concentrate to high PH levels. At high PH level, the tube steel's protective magnetic oxide coating is solubilized and rapid corrosion occurs as per the reaction given below:

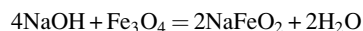
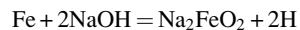




FIGURE 3.5

Hydrogen damage, “window opening” and “thick-edge spilt” type failure [3].

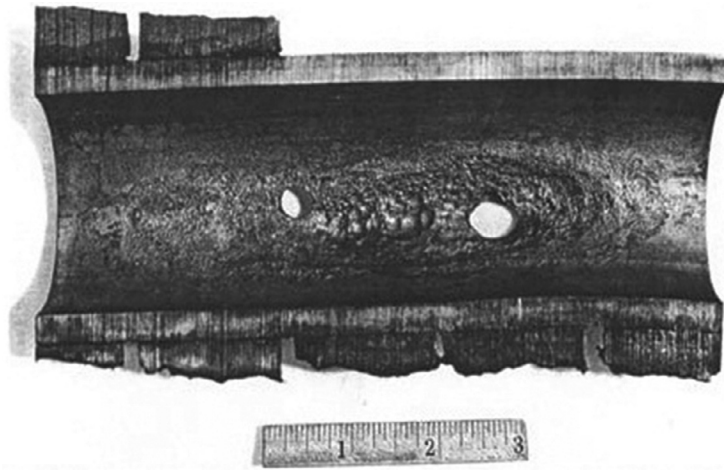
With the destruction of protective magnetic oxide layer, concentrated NaOH reacts with the tube material and forms atomic hydrogen as per the reaction



The atomic hydrogen so produced reacts with Fe_3C of pearlite constituent and forms CH_4 which ultimately causes hydrogen damage, discussed earlier. The tube surface deposits accumulate at locations where flow is disrupted such as, welds with backing rings, at bends, in horizontal tube weld, and at high-heat input locations. [Figure 3.2](#) shows the locations of the boiler where caustic corrosion can occur and [Figure 3.6](#) shows caustic corrosion gouging of a boiler tube.

2 CASE STUDIES

All possible types of boiler tube failure mechanisms have been discussed in detail earlier. Investigation into the probable causes of failure require gathering of physical evidence or data through a systematic problem analysis. The important

**FIGURE 3.6**

Caustic corrosion gouging. Severe localized gouging on the heat absorption side of this SA-210 waterwall tube was caused by caustic corrosion beneath a deposit. Once the caustic concentration within a deposit becomes grate enough to cause corrosion, the reaction becomes self-perpetuating and causes a through-wall leak within a very short time [3].

parameters that may require determination by calculation, observation, or measurement are as follows:

- Tube metal temperature
- Tube metal stress
- Tube metal thickness
- Tube metal diameter
- Tube metal material properties
- Tube metal microstructure
- Boiler water quality
- Boiler water flow
- Fuel constituents
- Fuel gas temperature
- Tube deposit constituents and
- Steam-side scale-thickness

2.1 CASE1: HIGH-TEMPERATURE CREEP

This investigation was conducted to determine the cause of failure of a platen superheater tube in a 140 MW thermal power plant. Visual examination along with dimensional measurement, chemical analysis, and hardness measurement were carried out to deduce the probable cause of failure. In addition, optical microscopy was a necessary supplement to understand the reason for the failure. It was concluded that the tube had failed due to severe creep damage. The tube metal had been exposed to a

temperature above the safe operating limit for a long duration. As a result, the tube metal had lost its strength, suffered creep deformation, and failed to support the working stress resulting in final rupture. The adherent scale, particularly, on internal surface formed due to exposure at high temperature. It had acted as an insulating blanket in view of its much lower thermal conductivity than that of the tube metal. The progressive build-up of scale had greatly affected heat transfer and raised the tube metal temperature above A1 before bursting of the tube.

Technical data relating to the problem are as follows:

- (i) Material specification: BS 3059/622-50 S2
- (ii) Dimension of original tube: 38 mm OD $\times$ 5.9 mm thickness
- (iii) Working temperature: 480 °C
- (iv) Working pressure: 158.2 kg/cm²
- (v) Running hours: 63,177

2.1.1 Examination details

2.1.1.1 Visual inspection

The tube sample was visually inspected and following physical evidences were noted.

- (i) The failure is open burst rupture. The general view of failed tube sample is shown below (Figure 3.7).
- (ii) Appreciable bulging/swelling at and adjacent to failure area.
- (iii) Reduction in wall thickness at the vicinity of rupture.
- (iv) Presence of thick black scale on the inner surface of the tube.
- (v) No significant evidence of corrosion on the outer surface of the tube.
- (vi) Presence of adherent black scale on the outer surface of the tube.

2.1.1.2 Dimensional measurement

Wall thickness measurement using thickness meter (type: DM-3, Krautkramer, Germany) and outside diameter measurements using vernier caliper were carried out on the tube sample. The locations of wall thickness and outside diameter measurement are shown in Figure 3.8 and the measured values are shown in Table 3.1.

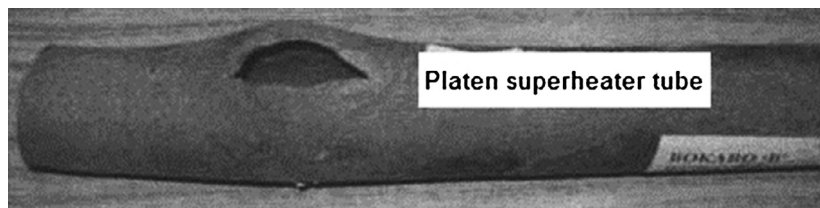
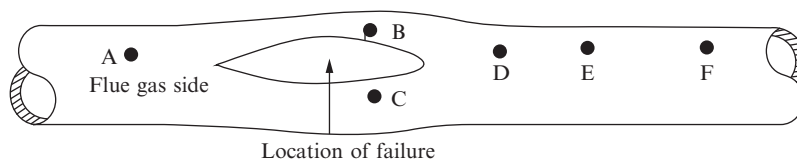


FIGURE 3.7

General view of burst of platen super heater tube.

**FIGURE 3.8**

Locations of wall thickness and outside diameter measurement of the tube sample.

Table 3.1 Measured Values of Wall Thickness and Outside Diameter of the Failed Tube Sample

Tube Identification	Location (Figure 3.8)	Measured Wall Thickness (mm)		Measured Outside Diameter (mm)	
		On Flue Gas Side	On Diametrically Opposite to Flue Gas Side	XX	YY
Platen superheater tube	A	5.2	6.1	39.62	41.1
	B	2.7	-	54.4	48.6
	C	2.3	-	-	-
	D	3.9	5.9	40.9	40.5
	E	5.5	6.3	39.5	40.0
	F	5.6	5.8	-	-

2.1.1.3 Chemical analysis

The material of the failed tube has been chemically analyzed and analytical results in percentage weight of the major constituent elements are incorporated in Table 3.2.

2.1.1.4 Metallographic examination and hardness measurement

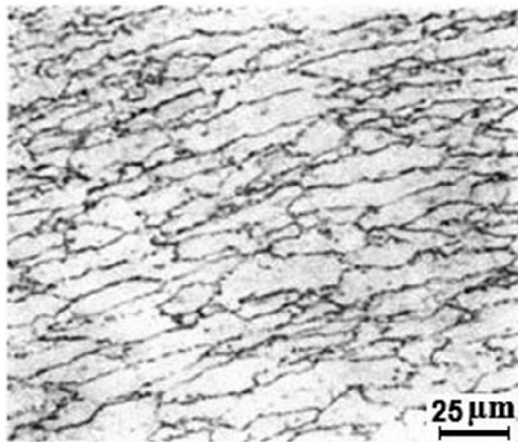
Metallographic specimens from the failed tube sample were selected, sectioned, polished, and etched as and when required as per the standard ASTM E-3, 1995, and ASTM E-407, 1993. All the specimens were examined under optical microscope for evaluating microstructural characteristics. The hardness measurements of the metallographic specimens were also carried out. The microstructural characteristics and hardness values are incorporated in Table 3.3.

Table 3.2 Chemical Composition

Tube Identification	Weight Percentage					
	C	Cr	Mo	Ni	Si	Mn
Platen superheater tube	0.09	2.10	0.90	-	-	-

Table 3.3 Microstructural Characteristics and Mean Hardness Values

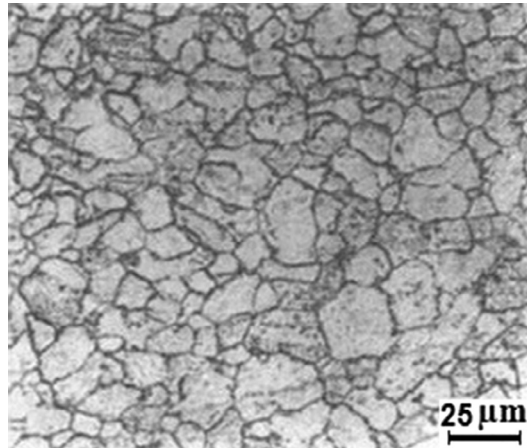
Sample	Location of Examination	Microstructural Characteristics	Mean Hardness Value (HV)
Platen superheater tube	Adjacent to failure	Elongated grains of ferrite and coalescence of alloy carbide at ferrite grain boundaries (Figure 3.9)	136
	Diametrically opposite to failure	Dispersed alloy carbide in ferrite matrix and coalescence of alloy carbide at ferrite grain boundaries along with isolated creep cavities (Figure 3.10)	128
	Away from failure	Ferrite, alloy carbide mostly at grain boundaries along with isolated creep cavities at the grain boundaries (Figure 3.11)	127

**FIGURE 3.9**

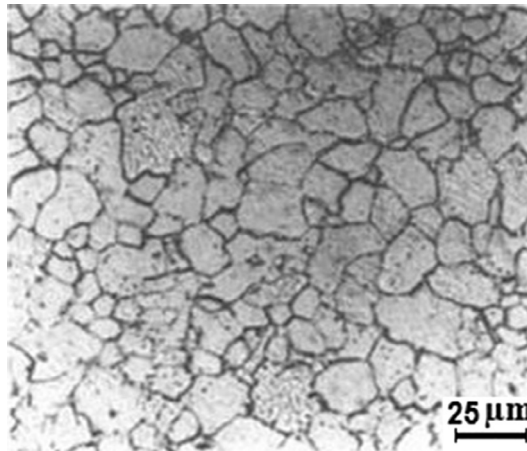
Micrograph showing elongated grains of ferrite and coalescence of alloy carbides.

2.1.1.5 Scale thickness measurement and estimation of tube metal temperature

Inside diameter scale-thickness measurement was carried out on three specimens from failed platen superheater tube. Tube metal temperature of the tube was estimated based on maximum measured oxide scale thickness on ID surface for 2¹/₄ Cr, 1 Mo steel. The observed maximum scale thickness and estimated tube metal temperatures are given in Table 3.4.

**FIGURE 3.10**

Ferrite and coalescence of alloy carbides along with isolated creep cavities.

**FIGURE 3.11**

Ferrite and alloy carbide at the grain boundaries along with isolated creep cavities.

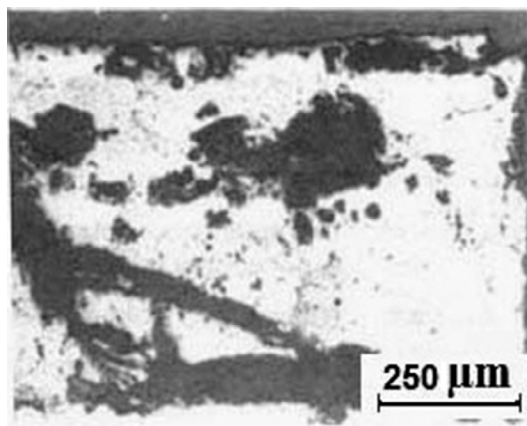
2.1.1.6 Discussion

The visual examination of the tube revealed that the tube had failed by a typical narrow open burst and the failed zone was accompanied by noticeable swelling of the tube circumference (Figure 3.7.). This indicates that the tube had suffered from noticeable creep deformation. Dimensional measurement of the tube revealed appreciable bulging/swelling and reduction in wall thickness at and adjacent to location of failure. A thick and tightly adherent scale on the internal surface over the length of

Table 3.4 Observed ID Scale Thickness and Mean Tube Metal Temperature of the Failed Platen Superheater Tube

Location of Scale Thickness Measurement	Maximum Measured ID Scale Thickness (mm)	Estimated Mean Tube Metal Temperature (°C)	Figure No.
(a) Adjacent to rupture of the failed tube	0.60	603	3.12
(b) Diametrically opposite to rupture	0.50	594	3.13
(c) Away from rupture	0.50	594	3.14

the failed tube was also observed (Figure 3.12). Steam-side adherent scale had resulted from steam-iron reaction which generally takes place in the temperature range of 450-700 °C [1]. The oxidation occurs either by inward diffusion of oxygen of the steam or by outward diffusion of iron. Microstructural features in the vicinity of the rupture included a mixture of elongated grains of ferrite and coalescence of alloy carbide at ferrite grain boundaries (Figure 3.9). On the other hand, microstructural features at the areas diametrically opposite to the rupture and away from rupture exhibited dispersed alloy carbide in ferrite matrix and coalescence of alloy carbide at ferrite grain boundaries along with isolated creep cavities (Figure 3.10). Elongated grains reveal that the material experienced high-temperature deformation at the location of rupture. The adherent scale, particularly on the internal surface, formed due to exposure at high temperature [2,11,12]. It had acted as an insulating blanket in view of

**FIGURE 3.12**

Tube inside oxide scale.

its much lower thermal conductivity than that of the tube metal. The progressive build-up of scale had greatly affected heat transfer and raised the tube metal temperature above A1 for a short duration before bursting of the tube. The estimated exposure temperature of the tube was approximated to be 603 °C, which is well above the intended operating temperature in the platen superheater zone [2]. The rise in temperature is the result of internal oxide layer which impeded the efficient heat transfer. The result was degradation of the tube material microstructure upon exposure to an elevated operating temperature over an extended period of time.

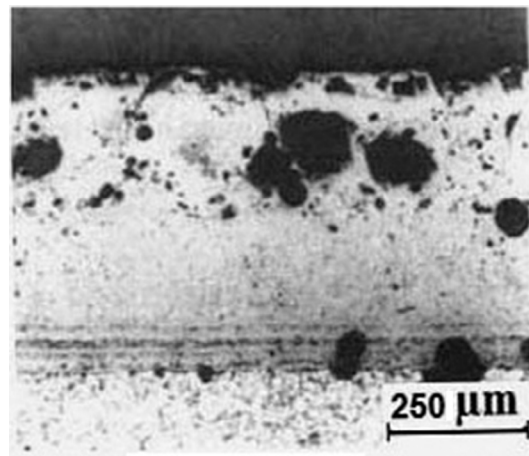


FIGURE 3.13

Tube inside oxide scale.

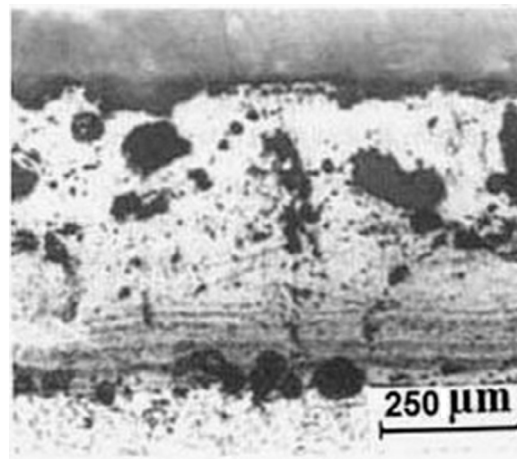


FIGURE 3.14

Tube inside oxide scale.

2.1.1.7 Conclusion for case study 1

Based on technical information provided and the findings of laboratory study, it is concluded that the tube had failed due to severe creep damage caused by high metal temperature during service. The probable causes of high metal temperature may be in sufficient flow of steam due to partial blockage, presence of thick oxide scale on ID surface, high flue gas temperature, etc.

2.2 CASE 2: HYDROGEN DAMAGE

This investigation was conducted to determine the cause of failure of a furnace wall tube in a 110 MW thermal power plant. Visual examination along with dimensional measurement, chemical analysis, and hardness measurement were carried out to predict the probable cause of failure. In addition, optical microscopy was a necessary supplement to understand the reason for the failure. It was concluded that the tube had failed due to severe hydrogen damage.

2.2.1 Examination details

2.2.1.1 Visual inspection

The tube sample was visually inspected and following physical evidences were noted.

- (i) The failure is thick lip wide-open burst ([Figure 3.15](#)).
- (ii) The maximum width of the failure is about 32 mm and the length of failure is about 200 mm.
- (iii) Noticeable bulging/swelling around the failure zone was observed.
- (iv) No appreciable wall thickness reduction around the failure zone.
- (v) No adherent deposits on inner side of the tube.

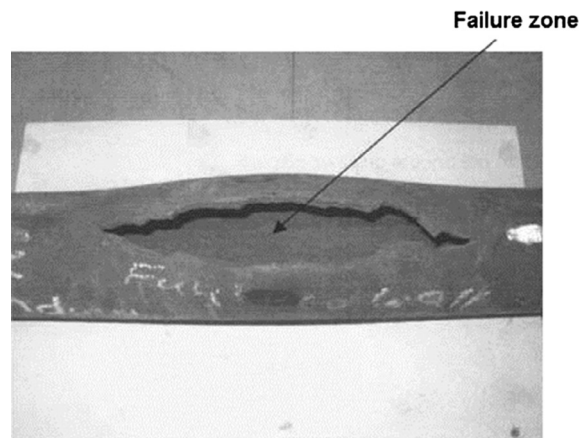
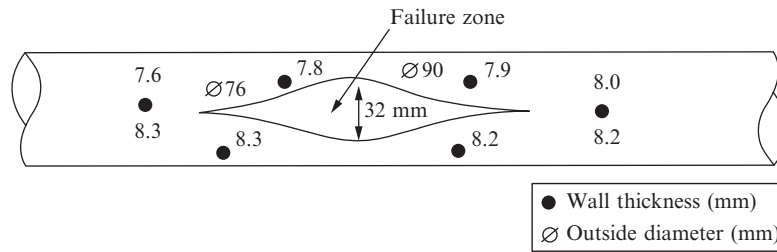


FIGURE 3.15

Failed furnace wall tube.

**FIGURE 3.16**

Locations of wall thickness and outside diameter measurement of tube sample.

2.2.1.2 Dimensional measurement

Wall thickness measurement using thickness meter (type: DM-3, Krautkramer, Germany) and outside diameter measurements using vernier caliper were carried out on the tube sample. The results are self-explanatory and are shown in [Figure 3.16](#).

2.2.1.3 Chemical analysis

Drilled chips from tube sample were carefully collected and chemically analyzed to estimate the weight percentage of the constituent elements. The chemical compositions are shown in [Table 3.5](#).

2.2.1.4 Metallographic examination and hardness measurement

Metallographic specimens from the failed tube sample were selected, sectioned, polished, and chemically etched as and when required as per the standard ASTM E-3, 1995, and ASTM E-407, 1993. All the specimens were examined under optical microscope for evaluating microstructural characteristics. The hardness measurements of the metallographic specimens were also carried out. The microstructural characteristics and hardness values are incorporated in [Table 3.6](#).

2.2.1.5 Discussion

Visual examination of the failed furnace wall tube revealed that the failure was a “thick lip” wide-open burst rupture associated with considerable bulging/swelling. Dimensional measurement revealed no considerable wall thickness reduction in and around the failure zone. The percentage increase in diameter was found to be 18% around the failure zone, which was evidenced by considerable bulging/swelling. The chemical composition of the tube conformed to the specification.

Table 3.5 Chemical Composition

Tube Identification	Weight Percentage		
	C	Si	Mn
Furnace wall tube	0.15	0.12	0.47

Table 3.6 Microstructural Characteristics and Mean Hardness Values

Tube Identification	Microstructural Features	Mean Hardness (Hv)
Adjacent to failure zone	(i) Discontinuous cracking (Figure 3.17) (ii) Discontinuous grain boundary cracking initiated from ID surface and presence of decarburization zone (Figure 3.18)	121.5
Slightly away from the failure zone	Ferrite, complete breakdown of pearlite in association with intergranular grain boundary cracking and decarburized zone (Figure 3.19)	-
Diametrically opposite to failure	Ferrite and partial breakdown of pearlite (Figure 3.20)	126
Away from the failed zone	Ferrite and pearlite (Figure 3.21)	124.5

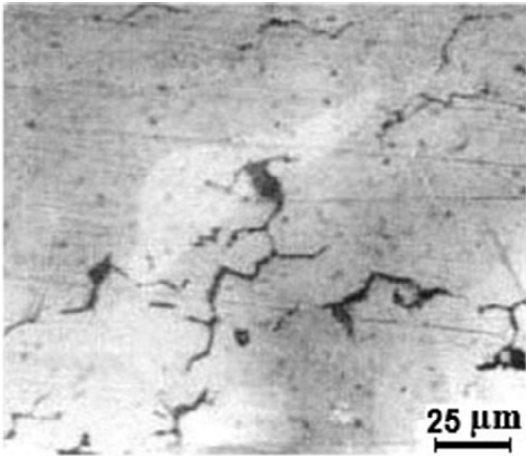
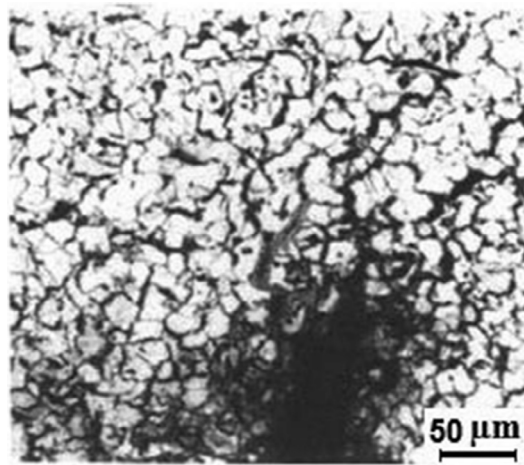
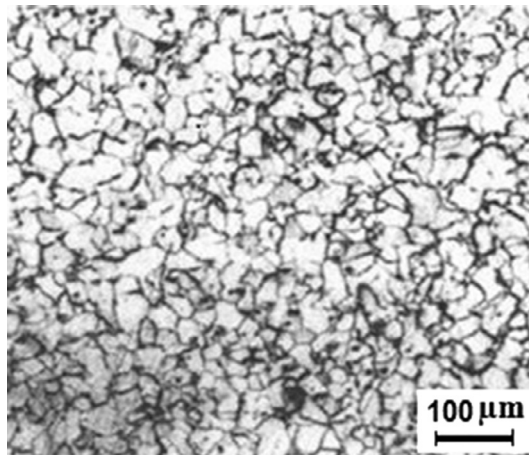


FIGURE 3.17
Micrograph showing discontinuous cracking.

The microstructural examination revealed ferrite, complete breakdown of pearlite in association with discontinuous intergranular cracking, and decarburized zone. This was probably due to hydrogen embrittlement. The carbon in the material reacted with the hydrogen to form methane (CH_4). Methane was a large molecule, which exerted pressure and caused discontinuous cracking along the grain boundaries [1,13,14]. Removal of the carbon from the sample caused the decarburization zone. The hardness measurement also showed no abnormal variation of hardness adjacent to the failure zone.

**FIGURE 3.18**

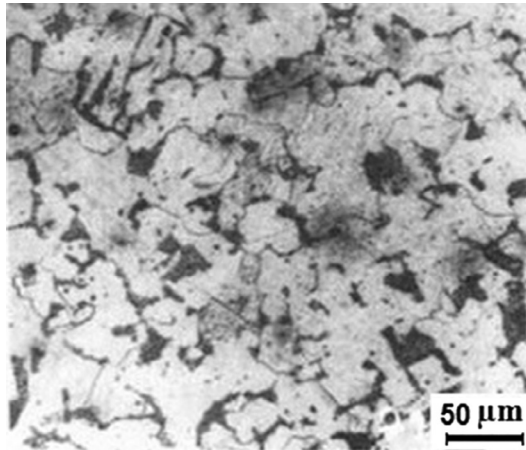
Micrograph showing discontinuous grain boundary cracking.

**FIGURE 3.19**

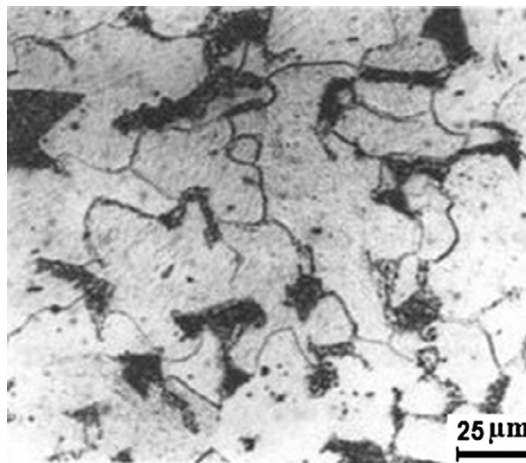
Micrograph showing ferrite and intergranular grain boundary.

2.2.1.6 Conclusion for case study 2

Hydrogen damage in the plain carbon steel normally occurs in high-heat flux region in the area of flow disruption such as welded joints with backing rings or protrusions, bends, or deposits. Hydrogen is caused by operation with low pH water chemistry from ingress of acidic salts through condenser leakage, contamination from chemical cleaning or malfunctioning of chemical control components, and concentration of

**FIGURE 3.20**

Micrograph showing ferrite and partial breakdown of pearlite.

**FIGURE 3.21**

Micrograph showing ferrite and pearlite.

corrosive contaminants within deposits on the inner tube wall. Deposits are formed from feed water system corrosion from condenser in-leakage constituent.

Corrective action involves the restoration of boiler water chemistry to the proper values and the consideration of boiler chemical cleaning. Chemical cleaning are performed to remove the internal deposits and to further generation of hydrogen on the tube surface.

3 CONCLUSION

Case studies reported indicated that actual mechanism of boiler tube failure can be identified through systematic metallurgical investigation. Based on the mechanism of failure, sequence of events that led to failure can be drawn and corrective actions can be suggested for preventing re-occurrence of similar failure. Hence, it may be concluded that investigation into the cause of boiler tube failure is very much helpful in improving the availability and reliability of boilers.

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Creep damage of high alloyed reformer tubes

4

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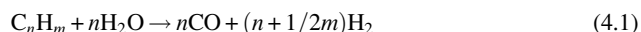
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1 INTRODUCTION

Creep-resistant tubes are used in a large number of industrial fields, but their main application is in petrochemical industry for hydrogen production by steam reforming [1].

This process takes place in furnaces, where a mixture of hydrocarbon and steam flows through hundreds of vertical tubes. Here, a Ni-based catalyst activates the endothermic reaction that transforms reagents into molecular hydrogen and carbon monoxide.



This reaction proceeds at a temperature range of 850-900 °C and under pressure of 3-3.5 MPa, according to kinetics that depends on process parameters, catalyst, and tube size [2].

Reforming tubes are very critical components being exposed to severe conditions for long time during service. Each tube is formed by two or three pieces, produced by centrifugal casting, machined, and butt welded. Tubes have usually inside diameter of 60-200 mm, wall thickness in the range 10-25 mm, and total length of 10-15 m and are designed for a nominal life of 100,000 h in service with a diameter straining of about 3%, at temperatures and internal pressures, respectively, up to 980 °C and 35 bar [3]. The operating temperatures, together with the hoop stress due to internal pressure, give rise to typical creep conditions: stress around 30 MPa and deformation rate of $\sim 10^{-10} \text{ s}^{-1}$.

At a first analysis, failure mechanism of the alloys utilized for reforming tubes can be summarized as nucleation of cavities and their evolution into microcracks and final cracks, which are qualitatively shown on a typical strain-time curve, together with the applicability fields of inspecting methods (Figure 4.1).

Although reformer tubes have a design life over than 10 years, premature failures are often observed. Their life is primarily limited by creep damage, driven by a combination of high service temperatures and hoop stresses that can evolve also into catastrophic failure [4].

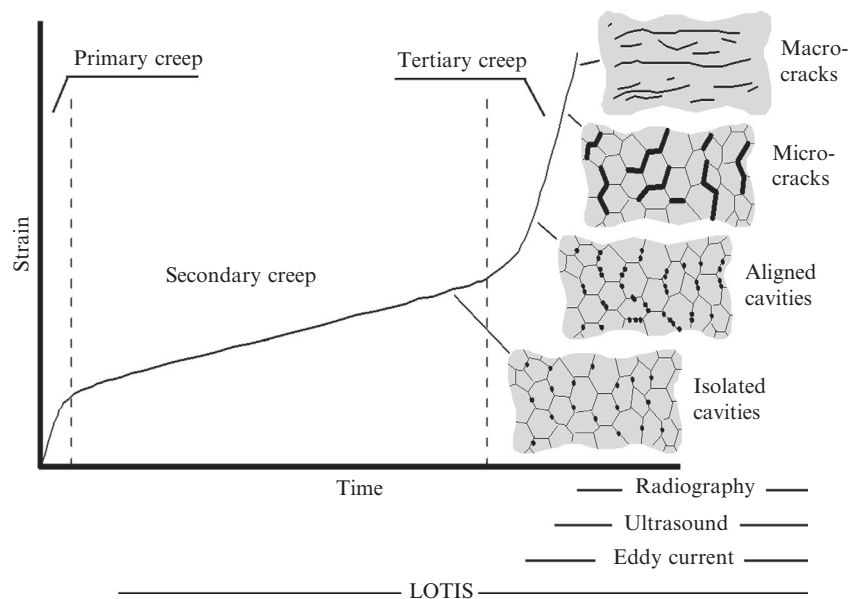
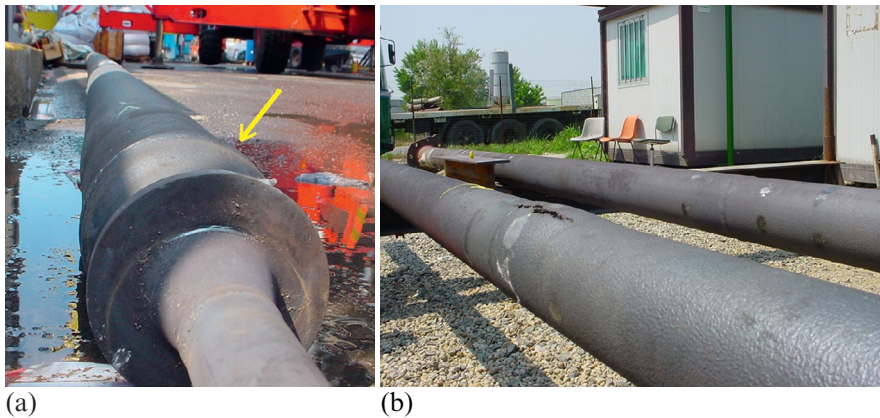


FIGURE 4.1

Typical creep curve and applicability fields of inspecting methods.

**FIGURE 4.2**

Reformer tubes put out of service: (a) diameter deformation indicated by the yellow arrow; (b) longitudinal crack in the deformed zone.

Creep life exhaustion is evidenced by changes in the tube alloy microstructure together with degradation of mechanical properties, giving rise to diameter elongations and longitudinal cracks after long-time service (Figure 4.2).

Cracks formation is the result of microstructure creep damage and thermal fatigue effects resulting from uneven temperature distribution in the tube wall (in both lengthwise and crosswise direction). Therefore, heat-resistant alloys with good mechanical properties and corrosion resistance are required. The cost of each tube is about 20,000 Euro, but the consequences of their sudden failure are far greater and difficult to predict.

1.1 CHEMICAL COMPOSITION AND MICROSTRUCTURAL FEATURES OF REFORMER TUBE

Tube materials have been widely improved over the last 50 years: formerly, in the 1960s and 1970s, the alloy mostly used was the 25Cr-20Ni-0.4C-Fe, designated as HK-40; during the following decades, the HP-40 grade (25Cr-35Ni-0.4C-Fe) with more Ni content has become common because of its better mechanical properties at high operating temperature [5].

The high nickel concentrations provide an austenitic microstructure giving, together with chromium, good mechanical strength and corrosion resistance up to service temperatures; moreover, the increase in Ni content, especially in conjunction with Si (>1%), improves the carburization resistance, a desirable property being the environment inside tubes moderately carburizing [6].

Due to their high mechanical strength, tubes are not easily drawn or extruded, thus cast structures are produced. Centrifugal casting is used to obtain a more even

austenitic microstructure that provides good mechanical strength and creep resistance according to its morphology [7].

Grain size and shape depend mainly on cooling rate during casting solidification. Generally, it can be stated that creep-resistant tubes reach the highest quality when coarse columnar dendrites, radially oriented, affect the entire cross-section of the casting wall [8].

Mechanical strength at high temperatures is ensured by a dispersion of hard deformation-resistant Cr carbide particles in the interdendritic regions, consequently creep behavior depends on their stability [9].

Unfortunately during long-term service at high temperature, material microstructure undergoes degradation phenomena because carbides changes their morphology tending to particles coarsening in order to minimize surface energy [10].

In literature [11] it was shown, through aging experiments, that after short-term permanence at temperature of 900 °C the material microstructure is generally subjected to degradation that accumulates as time progresses: carbon precipitation occurs in the early stages, then coalescence and coarsening of carbides result in embrittlement and reduction of strength. Last, further degradation [12] may lead to creep cavities, cavity coalescence, interdendritic microcracks, and final propagation of macrocracks during the tertiary stage of creep.

In this regard, an image of longitudinal cracks and an optical micrograph of tube wall, performed by us, are given in Figure 4.3: the observable cracks grow through internal microcracks that propagate following interdendritic paths, where carbide precipitation and microvoid nucleation occur.

Many researches have addressed their effort to optimize chemical composition and delay creep damage phenomena that occur inevitably: additional strengthening is provided by solid-solution of transition metals such as niobium in the HP-Nb alloy, where the primary NbC carbides have a lamellar morphology [13].

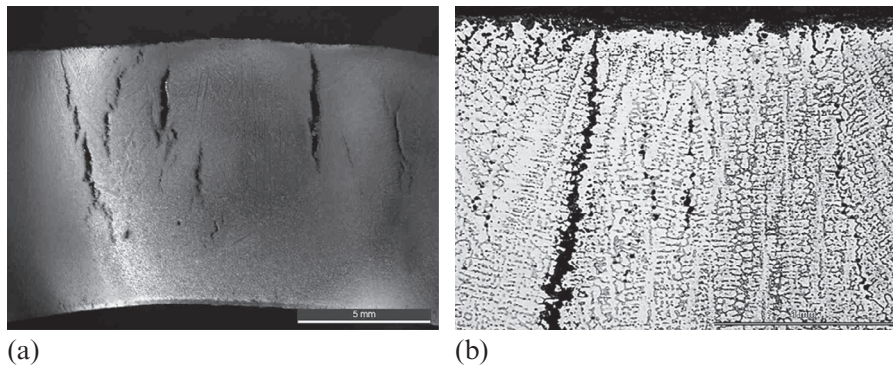


FIGURE 4.3

Image of longitudinal cracks (a) and optical micrograph of interdendritic macrocracks in the tube wall (b).

A new generation of reformer tube alloy began to be used in the 1980s, characterized by the presence of Nb, in the range 0.8-1.2%, as carbide-forming element stronger than Cr; moreover, the addition of cobalt and tungsten enhances the effects of both solution and precipitation hardening, improving creep resistance [14].

As pointed out in literature, a fine dispersion of carbide particles can be stabilized by modifying the HP grade composition adding stabilizing elements, such as the aforementioned niobium, titanium up to ~0.8% [15], or yttrium about 0.3% [16]. The outcome of this microalloying is a strong refinement of microprecipitates and improved carbides stability [14].

Starting from the early studies of Wen-Tai and Honeycombe [17], new austenitic alloys were developed adding, in addition to niobium, microalloying elements from groups III and IV of the periodic table (Ti, Ta, and Zr) and also rare earth elements (Cs, Nd, Pr, and Hf). These elements usually do not exceed a content of 0.5%, because it was experimented that when taken individually their effect on creep resistance may be contained, whereas taken jointly they show a synergic effect at lower concentrations and for given proportions. For example, the best combined effect of the joint addition of Nb and Ti is reached with a proportional share of their atomic contents equal to $Ti/(Ti + Nb) \approx 0.5$ [14].

1.2 PREDICTION OF CREEP LIFE BY MEANS OF THE LARSON-MILLER PARAMETER

Material selection for reformer tubes needs a complete understanding of the risk of probable failures caused by cracks within tubes during long-time service, as well as during the thermal cycles associated to the planned start-up and shut-down stages of plants.

Since the Second World War, materials researches in the field of heat-resistant steels and alloys were addressed to investigate creep properties, performing tests, and collecting data by public institutes and within industry.

Table 4.1 summarizes chemical composition and relative strength of heat-resistant alloys developed during the last 50 years.

Table 4.1 Heat-Resistant Alloys Compositions Developed During the Last 50 Years

Decade	Chemical Composition (wt%)							Relative Strength [14]
	Cr	Ni	C	Si	Nb	Other elem.	Fe	
1960s	25	20	0.2-0.6	1-2			bal.	1.0
1970s	25	24	0.2-0.6	1-2			bal.	1.4
1980s	23	22	0.2-0.6	1-2	~1	Ti	bal.	1.7
1990s to day	25	35	0.2-0.6	1-2	~1	Ti, Zr, W, and Cs	bal.	2.2

In order to estimate the long-term creep properties using results obtained over a short period, a number of time-temperature parameters have been developed. Among these, the Larson-Miller parameter (LMP) is the most commonly used around the world since the middle of the last century [18]. It is a means of predicting lifetime of materials using a correlative approach based on the Arrhenius relationship for thermal-activated processes.

Creep tests are carried out to evaluate the rupture times corresponding to the experimental values of temperature and applied load. The applied creep stresses (σ) are plotted versus the LMP values, expressed as follow:

$$\text{LMP} = T(C + \log t)/1000 \quad (4.2)$$

being T (°K) the test temperature, t (h) the rupture time, and C a material constant equal to ~ 20 for ferritic steels and ~ 15 for austenitic stainless steel. For the alloy under consideration, $C = 22.9$.

At a given stress, the LMP value is constant for a given material. So it is convenient to carry out accelerated creep tests at high temperatures to have short rupture times and calculate the corresponding LMP values: in this way, it is possible to predict very long rupture times at the specific service conditions that are characterized by the same applied stresses and lower temperatures than tests.

Obviously materials with the best creep behavior have, for a given applied stress, the greatest values of LMP. In Figure 4.4, the experimental σ versus LMP curves of

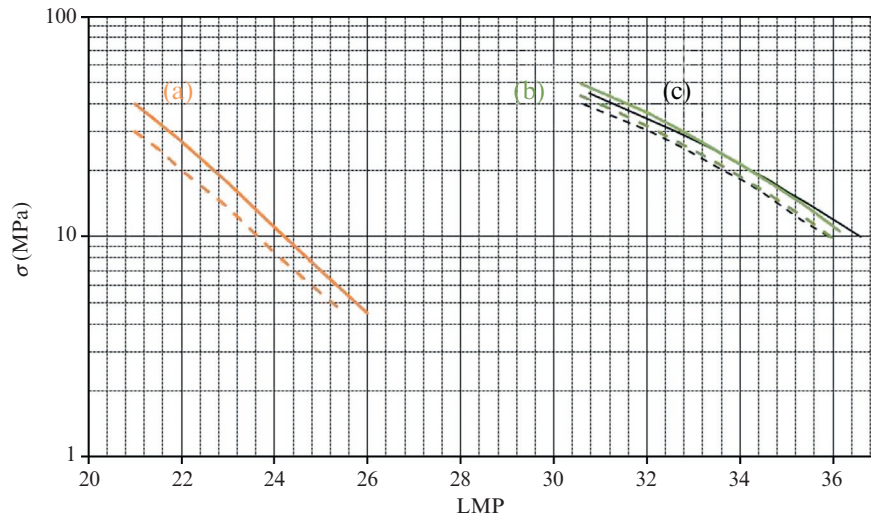


FIGURE 4.4

Applied stress versus LMP experimental diagrams, being (—) average values, (---) minimum values, (a) HK-40 grade [19], (b) HP-40 grade SEL2535Nb (Cronite) [20], (c) HP-40 grade G4852Micro [21].

an HK-40 and two HP-40 grades of two different manufactures are showed for comparison.

Entering in the diagram with a given σ level, the HP alloys show greater LMP values than the HK-40 grade.

The superior creep properties of the HP-40 alloys, compared to the HK-40 grade, provide longer rupture life at the same service temperature and allow to produce tubes with thinner wall or, alternatively, fewer tubes with larger diameter and save in tubing cost and furnace size. Moreover, it is worth to notice that the slightly different creep behavior of the two HP-40 alloys is probably due to the little differences of carbon and niobium contents that have percentages of 0.40 and 0.85, respectively, in the Cronite alloy and 0.45 and 1.5 in the G4852 Micro alloy.

1.3 A PROCEDURE FOR NONDESTRUCTIVE TESTING ON REFORMER TUBES

Creep properties depend on material microstructure and it is well known that tensile and creep strength decrease with the increase of tube working time, when changes in metallurgical features become clearly observable [22].

Investigations on these phenomena require to extract some tubes during planned maintenance shutdown of plant and then cut samples from them. Unfortunately, it must be taken into account that this procedure is really expensive, being the cost of each tube very high (some tens of thousands of Euros); moreover, the condition of a sample tube may or may not be representative of the total number of tubes in the furnace. Therefore, it seems worth to develop methods that allow to perform measurements on a large number of tubes, preferably without extracting them from plant.

Assessment on creep damage and residual life of reformer tubes has been addressed to monitor defects accumulation: as quoted in Ref. [23], conventional non-destructive testing techniques, such as eddy current and ultrasonic measurements, allow to detect internal cracks and are useful during the last service stage of tubes; material degradation phenomena, as carburization and oxidation, can be inspected observing ferromagnetic layers on tubes surfaces; measurements of magnetic permeability and low magnetic fields can give satisfactory information on effective tube wall thickness, but cannot indicate creep failures unless micro cracks appear.

In particular, a comprehensive approach should include reformer tube inspection, remaining life evaluation, and assessment of operational problems [19], as internal and external inspection methods to check periodically tube conditions at various creep stages and identify problems with burners or local hot spots, due to gas flow conditions, evaluating also the effect of high-temperature corrosion. Unfortunately, assessments of tube residual life are affected by several uncertainty factors.

As quoted previously, investigations on relationship between mechanical properties and microstructure degradation are current; however, reliable criterion describing the amount of degradation in creep-resistant tubes during service has not been established so far [24]. Experiments to predict residual life should be based on investigations performed during service, but up today they require the tube

decommissioning to perform metallographic examinations and mechanical tests in order to have a detailed description about creep damage.

It should be also underlined that an assessment of tube residual life is affected by several uncertainty factors, such as the actual value of operating temperature (temperature is measured by means of optical pyrometers with a measurement error of ± 20 °C); the reliable values of material creep properties (long-term properties for such alloys are only known by manufacturer catalogues) or the actual stress state acting on the tubes during service.

Moreover, in reformer tubes, thermal stresses during start-up and shut-down stages may be significant and lead to a creep-fatigue interaction damage mechanism [25]. As a consequence reliable means of inspection are required for a safety service of tubes.

Currently various methods, based on different types of piping inspection (Figure 4.1), are experimented by plant operators in order to develop a procedure for failure forecast and prevention [26,27]. In the last decade, reforming plants have relied on tube diameter deformation data, obtained *in situ* by the Laser Optic Tube Inspection System (LOTIS) [28].

Measurements of internal diameter (D_i) can be performed by inserting the LOTIS probe into a reformer tube by accessing through the upper flange. The probe, containing laser source, optics, and photo detector, is mechanically driven downward the tubes total length by an automated probe pusher and rotated up to 1800 rpm, depending on the desired resolution, so that a helical sampling pattern is generated during the tube internal surface mapping (Figure 4.5).

During the planned furnace shutdown, tubes can be inspected *in situ* by the LOTIS technique and put out of service when their deformation is judged not

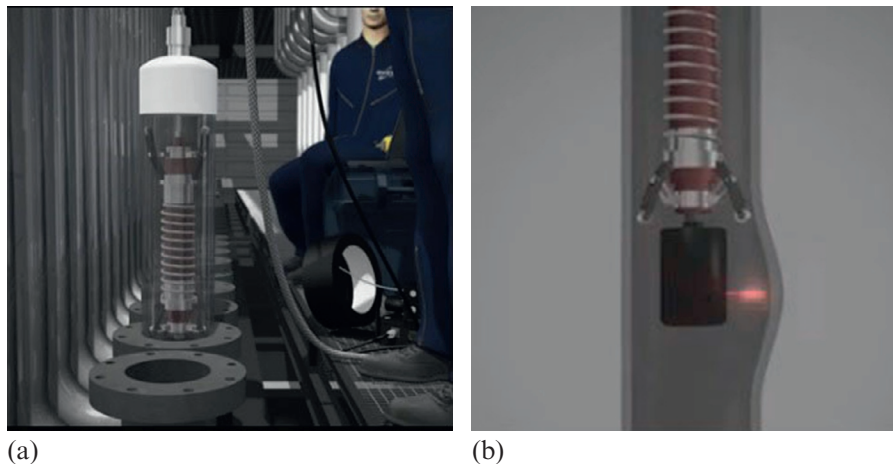


FIGURE 4.5

LOTIS measurements: (a) probe introduction into the reformer tube, (b) internal radius measurements by means of helical scans, recorded along the tube total length [29].

acceptable. In Refs. [30,31], it was argued that a deformation value of 1.5% can be assumed as a good criterion for deciding the tubes substitution.

The method for failure forecast and prevention, based on LOTIS diameter measurements, is reported here in order to show its reliability to determine the creep stage of material after long-time service and decide whether tubes need to be replaced. Therefore, metallurgical observations and creep tests were addressed to investigate creep damage of tubes made of a HP-40 alloy and analyze the adopted criterion for tubes decommissioning. In other words, it means verifying that diameter deformations lower than 1.5% are reasonably acceptable to ensure safe work until the next plant stop, that is usually planned every 2 years.

2 CASE HISTORY

The reformer furnace considered in this chapter is located in Sicily near the Tyrrhenian Sea (working today). The inspections reported in the following were carried out in 2010 on radiant tubes that had been in service for more than 10 years.

The furnace is a compact box, coated on the inside of refractory material, containing four parallel rows, each one with 44 vertical hanging radiant tubes, as showed by the axonometric sketch in [Figure 4.6](#). Rows of gas burners, installed at the furnace

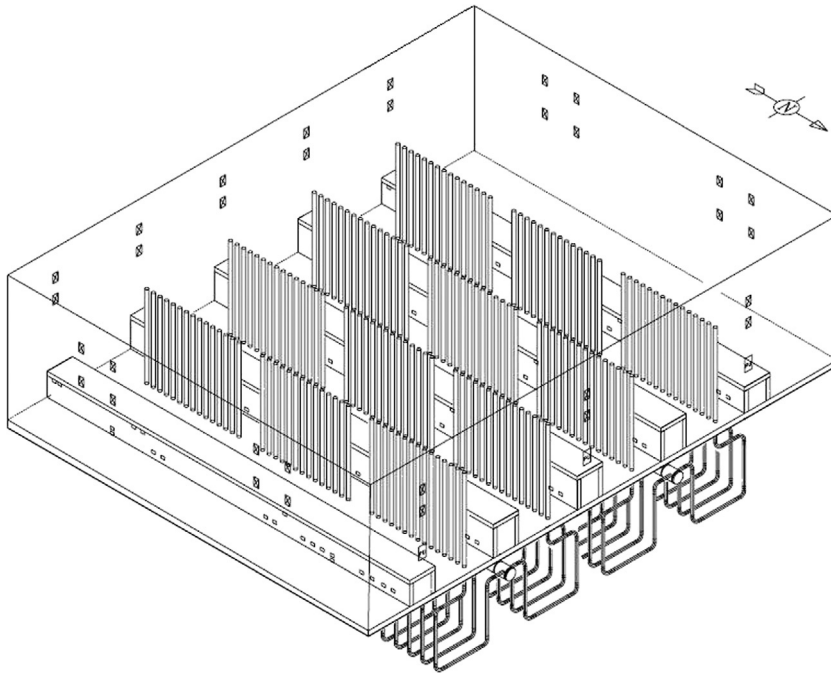


FIGURE 4.6

Axonometric sketch of reformer furnace.

top, supply energy for the endothermic reaction, heating hydrocarbon and steam from the inlet temperature ($\sim 500^\circ\text{C}$) up to the outlet temperature reached by the reaction products ($\sim 900^\circ\text{C}$).

The reformer tubes, heated from the outside, have nominal internal diameter of 101.6 mm, wall thickness of 10.5 mm, and length of 12.8 m; they were designed to work at 950°C with 32.8 bar of pressure for minimum 90,000 h and constructed in three pieces produced by centrifugal casting and then joined by gas tungsten arc welding (GTAW) (details on this procedure are given in [32]).

The investigated tubes are made of a modified HP grade alloy through the addition of Nb and Ti (Table 4.2), classified as ASTM 608 HP-Nb. The level of Ti and other microalloying element additions are generally proprietary and not disclosed by the alloy manufacturer.

These tubes are produced by centrifugal casting, so their microstructure consists of austenitic dendrites, oriented in radial direction and surrounded by eutectic carbides in the interdendritic regions.

2.1 EXPERIMENTAL METHODS

2.1.1 LOTIS measurements

Tubes deformation was evaluated by LOTIS technique during the plant shutdown that is planned every 2 years. Radius measurements of the tube inside surface were performed putting in rotation at high speed the probe containing laser source, optics and photo detector, while it was driven mechanically downward the tubes total length.

This laser-based surface mapping system provides an accuracy of 0.05 mm. Therefore, for a typical 100 mm internal diameter, it means that creep strain is quantified with a measuring accuracy of 0.05%.

On the basis of the internal diameter (D_i) measurements (mm), the deformation (ϵ) of a given tube section has been calculated with respect to the nominal internal diameter D_0 (equal to 101.6 mm):

$$\epsilon = (D_i - D_0)/D_0 \quad (4.3)$$

As discussed in Ref. [30], measurements of diameter deformation describe quantitatively creep damage. Tubes were decommissioned when their maximum diameter deformation exceeds the limit value $\epsilon = 1.5\%$, which does not ensure 2 years of safe work until the next plant stop.

2.1.2 Metallographic examinations and mechanical tests

Two types of samples for metallographic examinations and mechanical tests were investigated:

Table 4.2 Chemical Compositions Given by the Supplier (weight %)

Alloy	C	Si	Mn	Cr	Ni	Nb	Ti	Fe
A 608 HP-Nb	0.45	1.5	1	25	35	1.5	Add.	bal

- Type A: samples of the A 608 HP-Nb grade, taken from a tube in cast condition, as supplied.
- Type B: samples of the A 608 HP-Nb grade taken from segments with different values of the internal diameter, taken from tubes decommissioned after long-time service.

Tubes put out of service have an easy recognizable external expansion and surfaces strongly oxidized. In Figure 4.7, the degradation effects on surfaces are clearly observable on a segment cut from the zone of maximum diameter deformation.

With the aim of experimenting a correlation between creep expansion and material damage, and in order to confirm the validity of the tube substitution criterion, metallurgical observations, Vickers hardness, and creep tests were carried out on samples taken from different zones of decommissioned tubes. These zones were characterized by different values of the internal diameter.

Figure 4.8 shows where samples for metallurgical observations, Vickers hardness, and creep tests were cut. In particular, the dotted line ellipse highlights the zone where the maximum diameter deformation was recorded.

Creep accelerated tests were performed (according to EN ISO 204) on samples as supplied (type A), to compare the experimental data with the ones given by the manufacturer, and on samples (type B), cut longitudinally from tubes that were decommissioned, after 90,000 h (sample B1) or 114,000 h (sample B2/B6) of service.

All creep tests were carried out under constant load, with continuous monitoring of creep strains, with duration from 200 to 1000 h. Test temperatures were in the range 920-980°C, very close to design and service ones (950 and 900 °C,

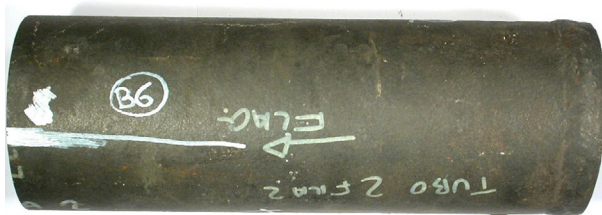


FIGURE 4.7

Tube segment cut from the zone of maximum diameter deformation.



FIGURE 4.8

Indications of the zones from which samples were cut.

respectively), while stresses were greater than design ones. In order to obtain short rupture time, test stresses were chosen greater than 17.5 MPa, that is the hoop stress corresponding to the design internal pressure of 32.8 bar.

Samples of tube wall were submitted to metallographic preparation by an automatic grinder-polisher system using a series of grit papers and a finishing diamond polish.

Metallurgical analysis was performed by optical and scanning electron microscopy (SEM). The utilized SEM apparatus was equipped with backscattered electrons detector and microanalysis device for energy dispersive X-ray analysis (EDX).

Samples were electrolytically etched by a solution of 10% oxalic acid for 8 s at 5 V before being examined by optical microscopy.

For SEM observations, samples were etched by a solution containing 15% of glycerol (100% of concentration), 45% of HNO_3 (65% of concentration), and 40% of HCl (37% of concentration).

2.2 RESULTS OF THE EXPERIMENTAL INVESTIGATIONS

2.2.1 Internal diameter measurements and hardness tests

Results of LOTIS measurements are given in Figure 4.9a by means of an internal diameter growth tube map of the reforming furnace. This map allows to choose the tubes to put out of service whether their internal diameter deformation exceeds the limit value $\epsilon = 1.5\%$, established for the plant under examination. Anyway, tubes' internal radius is not uniform along the tube axis (Figure 4.9b).

By way of example, results of LOTIS internal inspection along tubes, respectively, after 90,000 and 114,000 h, are given in Figure 4.10. The solid and the dashed line show that internal diameter is not uniform along the tube length: the upper segment of the tube results undeformed being the internal diameter equal to the nominal value, while the maximum internal diameter growth was recorded in the tube bottom where the highest temperatures are reached.

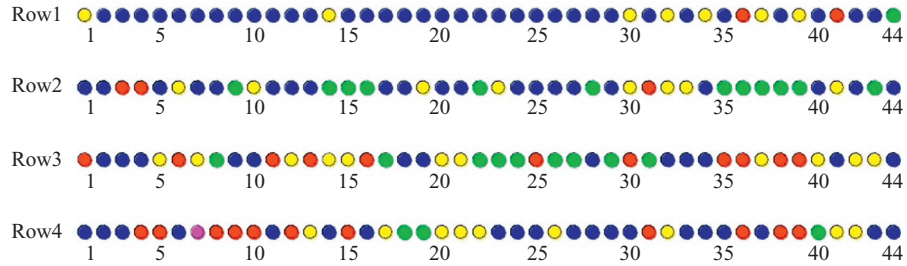
Vickers hardness results, performed across the thickness of different samples, cut from the tube decommissioned after 114,000 h (samples B2-B6, see Figure 4.8), were put in comparison with diameter deformation.

From Table 4.3, it can be noticed that at high diameter deformation corresponds small Vickers microhardness: samples B3 and B6 (cut from segments of the lower part of the tube where high diameter deformation was recorded) have smaller hardness values than the other samples and this can be considered as an effect of creep damage.

2.2.2 Metallurgical analysis—optical microscopy observations

Microscope observations were carried out on metallographic sections taken transversally to the tube wall. As pointed out in Refs. [30,31], metallographic features of tubes are characterized by austenitic dendritic grains oriented in radial direction (Figure 4.11), providing greater strength and better creep resistance to the circumferential stress generated by the inner pressure.

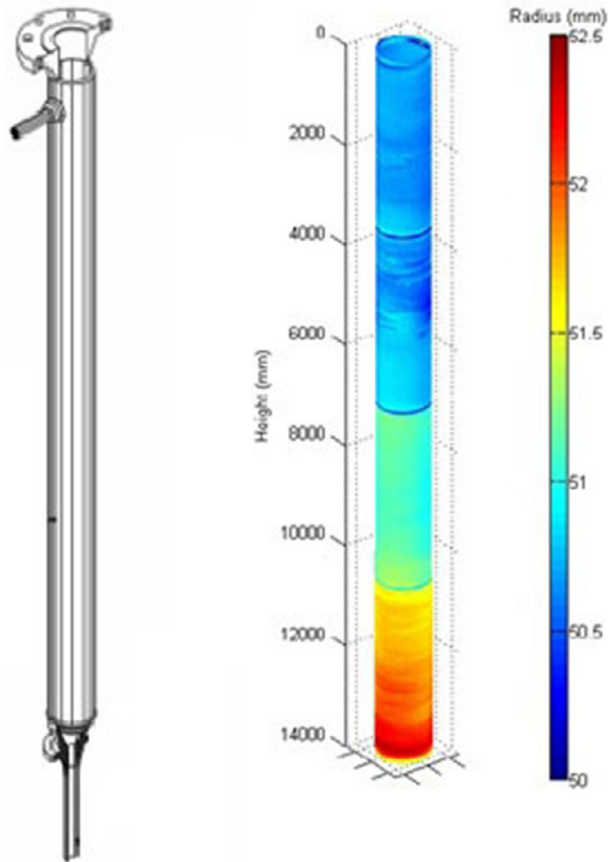
Internal diameter growth tube map



ID Growth legend

- <1.0%
- =1.0% < 1.5%
- =1.5% < 2.0%
- =2.0% < 3.0%
- >3.0%

(a)



(b)

FIGURE 4.9

Results of the LOTIS measurements: (a) internal diameter maximum deformation for every tube of the reforming furnace, (b) internal diameter deformation along a tube.

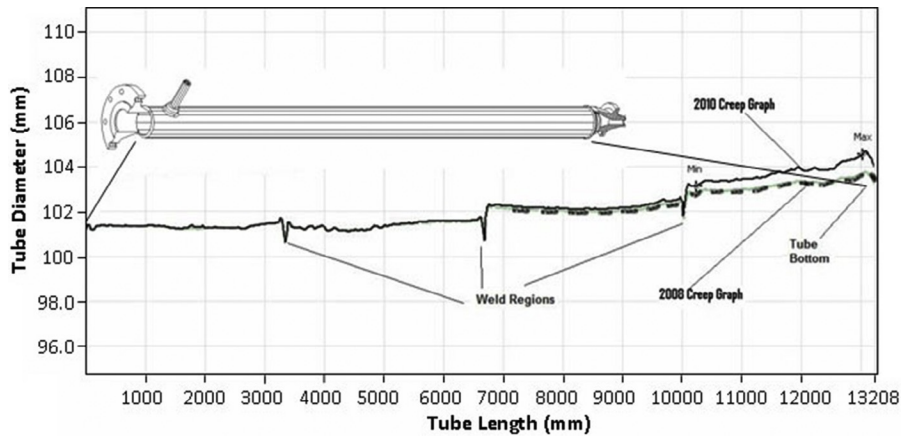


FIGURE 4.10

Internal diameter values versus tube length: creep graphs after 114,000 h (—) and after 90,000 h (---) [30].

Table 4.3 Internal Diameter and Hardness Values Measured on Samples B2-B6 (Tube Decommissioned After 114,000 h of Service)

Sample	Internal Diameter	Deformation ϵ (%) (See Equation 4.3)	Hardness (HV) (Mean Value Across the Thickness)
B2	102.4	0.8%	175
B3	103.5	1.9%	151
B4	100.8	0.0%	171
B5	102.2	0.6%	179
B6	104.0	2.4%	162

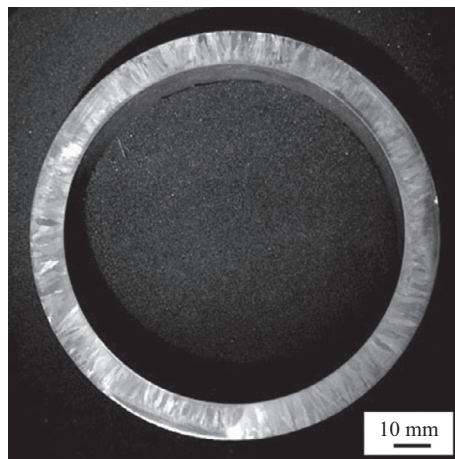


FIGURE 4.11

Optical metallography of a tube section.

Carbides are clearly observable, also at lower magnification, in the interdendritic region where they perform the role of enhancing creep strength.

After long-time service, this microstructure keeps only the dendritic features, because at higher magnification some important microstructural changes can be observed in the tube sections where high diameter deformation was recorded.

In [Figure 4.12](#), the microstructures in the as-supplied conditions (sample A) and in different conditions of creep deformation after long-time service (type B samples) are showed for comparison.

The as-cast microstructure is characterized by a continuous network of primary carbides, lying mainly along the austenitic dendritic grains boundary, and by decreasing Cr concentration in the matrix.

Service at high temperature lead to some microstructural changes, such as carbide coarsening at grain boundary, very fine precipitation of secondary carbides inside the austenitic matrix, and some isolated circular microvoids at grain boundary. The boundary coarsening is due to the high temperature and can be observed also in samples treated by aging at 900 °C for 2000 h [11].

Nevertheless, the last two phenomena are specific of samples B3 and B6, highly deformed after long-time service. Anyway, the absence of large peanut shape voids, formed by the coalesce of two or more voids when creep is in progress [9], can be assumed as an indication that the creep final stage is not yet started.

2.2.3 Metallurgical analysis—SEM observations

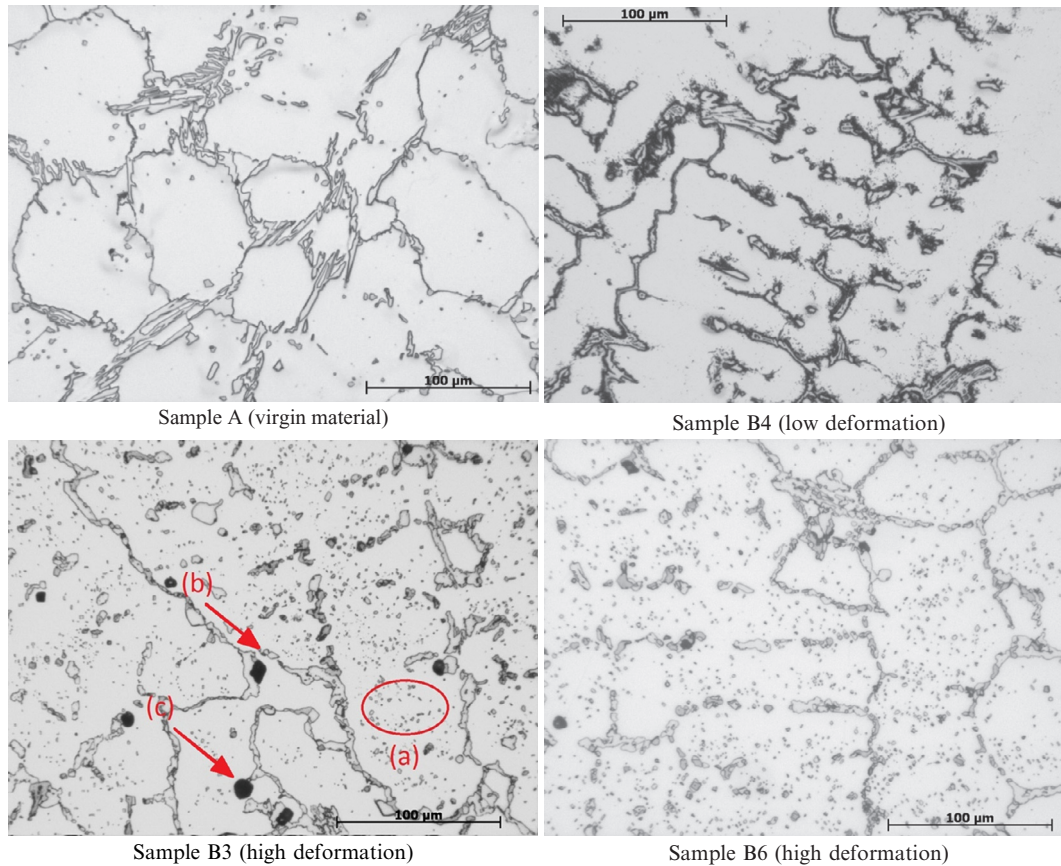
SEM images of the tube wall microstructure are given in [Figure 4.13](#). In the as-cast conditions (sample A) microstructure shows clearly a matrix of austenitic grains and two secondary phases. In fact, two different precipitates can be recognized for their different morphology: a dark gray phase located at grain boundaries and a phase in the form of small white particles. The dark gray phase has a volumetric fraction greater than the white particles.

The EDX spectra allow the following phases identification:

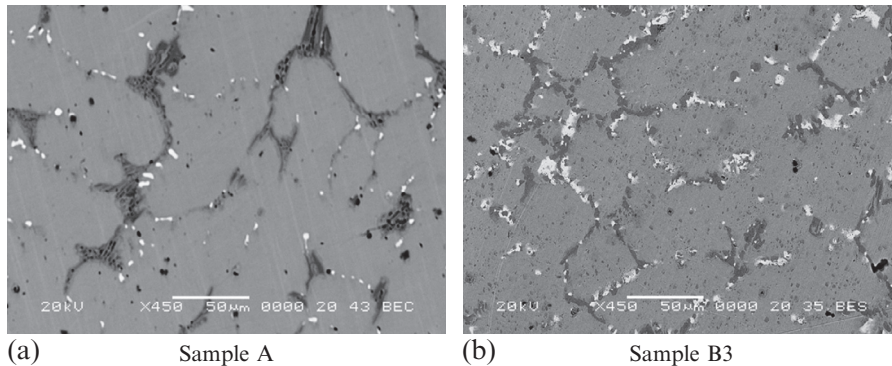
- an austenitic matrix with high content of Cr, Ni, and Fe;
- a dark phase very rich in only Cr and C that can be recognized as Cr carbide;
- a precipitate, white particles, with more concentration of Nb and Si and less content of the other elements with respect to the nominal composition.

As concerned as B3 sample (the sample having higher deformation), a dark gray precipitate can be observed both inside grains and at their boundaries.

The EDX measurements confirm the dark gray phase at grain boundaries as Cr carbide; also the dark gray small particles inside grains, very rich of Cr, are recognized as carbides. In fact HP steels, which are usually used in the as-cast condition, undergo phase transformations due to thermal activation during service at temperatures between 1123 and 1325 K: primary Cr carbides M_7C_3 are instable at high temperature and thus are transformed into intragranular and intergranular $M_{23}C_6$ precipitates [22].

**FIGURE 4.12**

Optical micrographs at high magnification of samples in the as-cast conditions (type A) and after long-term service (type B), taken from tube segments with different diameter deformations and corresponding degradation. Sample B4 shows negligible degradation due to service, while samples B3 and B6 show a fine carbide precipitation inside grains (a) primary carbide at grain boundary (b) and almost circular microvoids (c).

**FIGURE 4.13**

SEM micrographs at high magnification of samples in the as-cast conditions (sample A) and after long-term service (sample B3).

The white phase is in the form of precipitates at grain boundary of greater dimension than white particles in type A sample. Moreover, it is worth noting that the type B sample has a volumetric fraction of white precipitates greater than type A sample. In particular, the EDX measurements show that the white precipitates have a high concentration of Nb, Ni, and Si.

As quoted in literature [33,34], Nb carbides are not stable in the range of service temperatures, between 700 and 900 °C, being transformed into Ni-Nb silicates known as G-phase. So the Nb-rich particles can be ascribed to the G-phase formation, with the presence of intermetallic compounds in addition to Nb carbides which are not yet transformed.

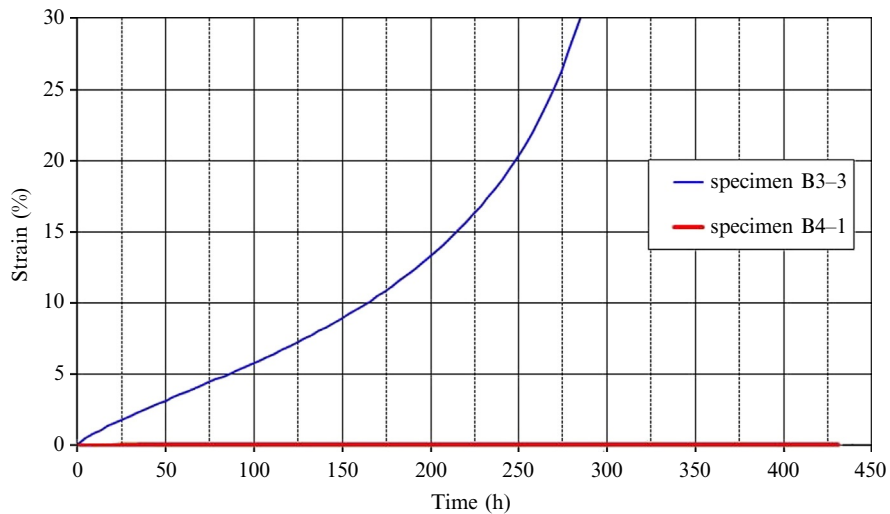
The increasing of the Nb-rich phase volume fraction follows the exposure at high temperature. It's worth considering that the Nb carbides give good creep properties, but the resistance to creep deformation comes down along with formation and coarsening of secondary phases [3].

Moreover, the transformation of these carbides into the G-phase has negative effects because the interface between G-phase and matrix is a preferential site for the occurrence of creep damage [7]. Luckily, this phenomenon can be overlooked in the HP grade microalloyed with small quantities of Ti that reduce the G-phase volume fraction [7].

SEM observations allow to document deterioration phenomena that in the investigated radiant tube affect only a little fraction of the wall thickness, at both the inner and outer surface. The SEM micrographs in [30] show damaged zones with a thickness of some tenths of microns in the tube wall.

2.2.4 Creep test and residual life prediction

Two creep curves are given for comparison in Figure 4.14: they refer to the B4-1 and B3-3 samples and were obtained in the same experimental conditions (temperature 980 °C, stress 23 MPa).

**FIGURE 4.14**

Creep curves for the B3-3 (from a deformed zone) and B4-1 (from an undeformed zone) samples.

Results of creep surveys are summarized in Table 4.4. In agreement with metallurgical analysis, the B3 samples, that were taken from a deformed zone characterized by considerable microstructural changes, have short creep life and high creep ductility. On the contrary, the B4 samples, cut from an undeformed section, exhibit creep behaviors very close to the as-supplied alloy (A specimen).

In order to compare the creep test results obtained in different experimental conditions, they are plotted (Figure 4.15) on a Larson-Miller diagram, where the material constant C [35] was assumed to be equal to 22.9 [30,31].

Table 4.4 Results of Creep Tests for Samples in the As-Cast Conditions (Type A) and After Long-Term Service (Type B), Taken from Zones with Different Values of Diameter Deformation

Samples	Condition	T (°C)	Applied Stress σ (MPa)	Rupture Time t (h)	Creep Ductility (%)
A	As cast	950	38.5	583	7.3
B4-1	Sample with low diameter deformation	980	23.0	>3000	-
B4-2		950	38.5	457	10.2
B3-1	Sample with high diameter deformation ($\epsilon > 1.5\%$)	950	25.6	366	41.9
B3-2		920	30.0	371	41.2
B3-3		980	23.0	285	30.9
B3-4		920	27.0	926	39.7

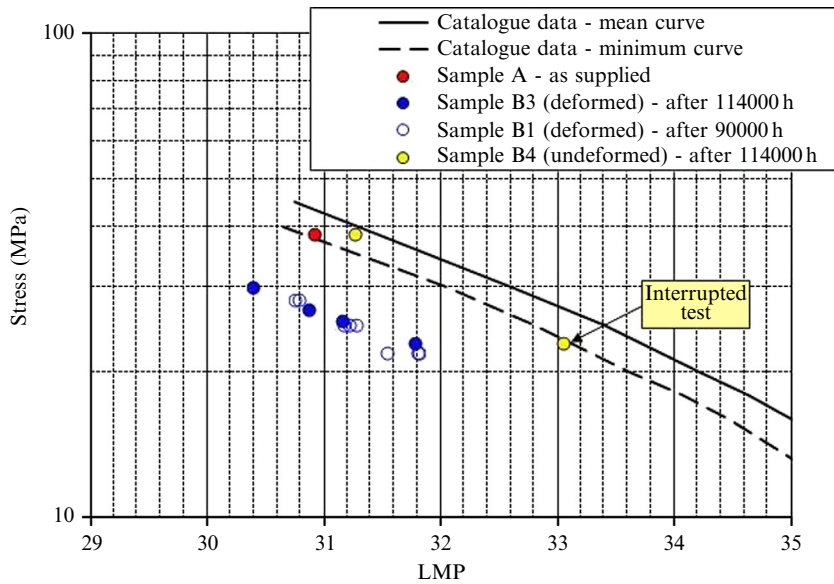


FIGURE 4.15

Larson miller diagram ($C=22.9$): experimental creep data for samples type A (as supplied) and type B after 90,000 h and 114,000 h of service. Larson-Miller curves supplied by the manufacturer catalogue [21]: average values (—) and minimum values (---).

Available creep test data performed on a tube of the same plant (sample B1), decommissioned after 90,000 h of service with maximum diameter deformation around 1.5% are also plotted.

On the basis of creep test results, the following considerations can be drawn:

- creep test result for type A specimen (as supplied) are consistent with the manufacturer minimum creep values;
- creep test results for B4-1 and B4-2 samples (taken after service from undeformed sections) are very close to the “as-supplied” behavior;
- the representative points of specimens (B3) and (B1), taken after service from deformed sections, are well below the manufacturer creep curves, due to creep damage accumulated during service;
- creep test results of samples taken from deformed sections, after 114,000 h of service (B3), are well aligned with creep test results obtained from similar sections of tubes removed from service after 90,000 h (B1);
- creep ductility is quite low for the “as-supplied” type A sample (7%), while it is higher for type B3 and B1 samples (nearly 40%).

As described in Ref. [30], the tube residual life was estimated by means of an experimental master curve, obtained correlating creep test results of B3 and B1 samples. Therefore, assuming $\sigma = 17.5$ MPa (hoop stress at design pressure), the LMP values (Figure 4.15) and then the corresponding rupture time at different temperatures

Table 4.5 Residual Life t_{res} of Reformer Tubes Without any Safety Factor (Experimental Evaluation)

T (°C)	t_{res} (h)	t_{res} (Years)
900	88,500	10.1
910	49,950	5.7
920	28,350	3.2
930	16,300	1.9
940	9400	1.1

(Equation 4.2) can be deduced. This rupture time can be considered as the tube residual life (t_{res}) in the case of keeping in service.

From Table 4.5, it is evident that the tube residual life is very sensitive to even small differences of temperature: changes of 10-20 °C inside the tube wall, which are in the range of measurement errors, may results in significant reductions of creep life.

A theoretical calculation of the residual life can be obtained simply by the difference between expected rupture time (given by the catalogue minimum curve) and effective service time. The theoretical residual life is compared with the experimental evaluation of residual life, based on the creep data of deformed samples (type B1 and B3 samples).

This calculation was performed assuming a pressure of 3.28 MPa (design pressure) and a temperature of 900 °C (operating temperature).

Results, given in Table 4.6, show that the experimental residual life is significantly lower than the residual life obtained by theoretical calculation.

This reduction of life is a direct consequence of creep damage accumulation. Moreover, the experimental residual life drop dramatically when a safety factor equal to 0.8 is applied on stress, in accordance with the Italian Standards.

Table 4.6 Residual (Life t_{res}) of Reformer Tubes (Comparison Between Theoretical and Experimental Evaluation Based on Creep Data of Samples Taken from Deformed Zones of Tube)

	Service Time (h)	Expected Rupture Time ^a (h)	Residual Life (h)	Residual Life ^b (%)
Theoretical calculation (minimum catalogue curve, safety factor $S=0.8$)	114,800	325,800	211,000	65
Experimental evaluation (no safety factor)	114,800	325,800	88,500	27
Experimental evaluation (safety factor $S=0.8$)	114,800	325,800	17,750	5

^aFrom catalogue minimum curve.

^bFraction of the expected rupture time.

Data of [Tables 4.5](#) and [4.6](#) show that by applying a safety factor the experimental residual life of tubes with diameter deformation greater than 1.5% become dangerously close to 2 years (the service time planned until the next plant stop), or even shorter in the case of some tens of centigrade higher.

3 CONCLUSION

Radiant tubes are very critical components being exposed to severe conditions for long time during service and, even though they are designed to work over than 10 years, premature failures may be often observed. Their life is primarily limited by creep damage, driven by a combination of high service temperatures and hoop stresses that can evolve also into catastrophic failure. In order to enhance creep behavior, materials have been widely improved from the first high Cr-Ni alloyed steel to the last generation of Cr-Ni alloys, in which microalloying elements have significant effects.

The tubes considered in this work, made of a HP grade alloy modified with Nb and Ti additions, were inspected *in situ*, during the plant stop planned every 2 years, through diameter measurements along their axis performed by a LOTIS. They were put out of service when diameter deformation reached a limit value of 1.5%. In order to assess the reliability of such criterion, tube's degradation was investigated to verify any correlations among diameter deformation, microstructural damage, and creep behavior.

Starting from the observation that the measured diameter values are not uniform along tube axis, it was ascertained that samples taken from undeformed sections have microstructural characteristics similar to those of samples in the as-cast conditions; while a significant degradation is present in samples taken from zones where diameter deformation exceeds the limit value assumed for tube decommissioning. In these zones, the Nb-rich phase transformed, during service, from Nb carbides into Ni-Nb silicates, known as G-phase, producing negative effects on creep behavior; but the large peanut shape voids, that give rise to creep failure, are not observed.

Concerning creep test results, samples taken from undeformed sections have the same behavior recorded in the as-cast conditions; differently a remarkable shortening of creep rupture time compared to the as-cast conditions is recorded in samples taken from deformed zones.

Therefore, from the results of these investigations, a link between tube diameter deformation and creep damage is evident. Moreover, considering the results of residual life calculation based on creep experimental data, it can be stated that the limit value of 1.5% for the diameter growth appears reasonably a conservative criterion for deciding the tubes substitution.

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Failure analysis of reinforced concrete structures subjected to chloride penetration and reinforcements corrosion

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1 INTRODUCTION

Concrete structures, especially reinforced concrete structures are one of most common type of structures used around world. The coupling between steel and concrete allows engineers to design composite structures considering complex architectural geometries with fair costs. When these structures are located in nonaggressive environments, they respect, in general, the structural life predicted. However, the structural durability is strongly reduced by environmental processes, which have physical, chemical, or mechanical origins [1–4].

In recent past decades, structural problems related to durability failure of reinforced concrete structures have increased drastically [5–7]. Figure 5.1 presents two columns mechanically damaged due to reinforcements corrosion effects. In this figure, crack growth and spalling phenomena are observed. In addition to economic and financial impacts due to repair costs, the durability problems also have industrial, environmental, and social consequences, due to the application of repair materials and the decrease of safety of these structures [8–10].

The high costs involved in both steel loss, structural material mechanical degradation and repair, the loss on structural stability and the recurrence of these problems make the corrosion of reinforcements the principal pathological manifestation in reinforced concrete structures [11–14]. Table 5.1 presents the cost involved into



FIGURE 5.1

Columns damaged due to reinforcements corrosion effects. Author's personal picture.

Table 5.1 Cost with Construction and Repair for Some European Countries [13]

Country	Cost with New Constructions (10^6 €)	Cost with Maintenance and Repair (10^6 €)
France	85.6	79.6
Germany	99.7	99.0
Italy	58.6	76.8
United Kingdom	60.7	61.2

construction and repair/maintenance in some countries in Europe. As observed by the numbers presented in this table, the cost associated with repair/maintenance cannot be disregarded.

The corrosion of reinforcements is directly connected to the durability of reinforced concrete structures. In addition to that, the diffusion of chlorides is recognized as one of most important sources that triggers the corrosion process [15,16]. Therefore, at modeling this phenomenon, the loss of reinforcements, the concrete damage level and, consequently, the structural durability are efficiently evaluated.

The chloride penetration into concrete pores is controlled by complex physical and chemical mechanisms. The modeling of this phenomenon is often simplified, without significant loss of representation, by a process controlled only by the diffusion of chloride ions into concrete pores. The corrosion starts when a threshold level of chloride concentration is reached at the vicinity of reinforcements, leading to its depassivation [17,18].

Several models have been proposed in literature to represent accurately such phenomenon. Among them, it is worth to mention [19,20] where Nernst-Planck-Poisson equation was applied to describe the movement of multispecies in saturated concrete. Bastidas-Arteaga et al. [21] presented a degradation model for reinforced concrete structures where mechanical degradation processes are caused by biodeterioration sources (i.e., action of live organisms), steel corrosion, and concrete cracking. Zhao et al. [22] analyzed the concrete cover damage induced by corrosion in reinforced concrete structures using on a simple analytical model. This model was based on damage and elastostatic mechanics, which considers noncracking and partial cracking stages.

Cellular automata approach was used to model the diffusion mechanism into concrete pores by Biondini et al. [23]. The mechanical damage coupled to diffusion is evaluated by introducing a mechanical degradation law for both concrete matrix and steel rebar in terms of suitable damage indices. The mechanical model proposed by Biondini et al. [23] was applied to assessment of reinforced concrete structures durability taking into account the inherent randomness on the problem variables by Biondini et al. [24]. The probabilistic analyses were performed using Monte Carlo simulation and structural maintenance planning based on reliability index were proposed. A case study in cable-stayed bridges using this model is presented in Ref. [25]. The evolution of the lifetime structural performance considering uncertainties is presented in Ref. [26] using the approaches introduced in Ref. [23–25].

The time-variant reliability analyses of reinforced concrete highway girder bridges subjected to time-dependent load and resistance is presented in Refs. [27–29]. A simplified mechanical model, which is based on an analytical equation and two random variables, is used to represent the girders mechanical behavior along time. In these works, Fick's law is adopted to determine the time for corrosion initiation and Monte Carlo simulation applied to calculate the probability of structural failure. Fick's law and simplified relations to represent the steel loss along time are adopted in Ref. [30] to formulate a reliability-based design optimization problem. The total structural cost is minimized taking into account construction and failure individual cost.

Some numerical studies using finite element method (FEM) were already presented in literature [31–33]. These studies analyze the mechanical problem of reinforcements corrosion penalizing the reinforcement area and modeling the concrete damage by empirical/analytical approaches. The researches cited above presented theoretical/numerical/empirical approaches that allow the mechanical modeling of reinforced concrete structures subjected to reinforcements corrosion. Nevertheless, a numerical approach that accounts realistically for both steel and concrete mechanical degradation along time coupled to a numerical method capable to general structural modeling is still a challenge in this important scientific domain.

In this regard, a numerical mechanical model for structural analysis of reinforced concrete frames subjected to chloride penetration and reinforcements corrosion based on FEM is proposed in this chapter. The nonlinear mechanical behavior of steel and concrete are modeled considering elastoplastic criterion and damage mechanics approach, respectively. Geometric nonlinearity is considered by the updated Lagrangian description, which allows writing the structural equilibrium conditions on the last equilibrium configuration. To improve the modeling of shear influence, concrete strength complementary mechanisms, such as aggregate interlock and dowel action are accounted. It is worth to mention that the modeling of such nonlinear phenomena is one contribution of this study. Fick's law is adopted to determine the chloride concentration growth at the structural cover along time. The loss of rebar's area along time due to corrosion process is calculated using the model proposed by Val and Stewart [17], Val and Melchers [34], and Vu and Stewart [35], which is based on experimental data.

In this study, bended structures are analyzed considering the proposed nonlinear numerical formulation. The results provided by the proposed model are compared with responses available in literature. The proposed nonlinear FEM formulation allows the assessment of the residual structural resistance along time accounting the chloride penetration and the structural loss resistance due to corrosion phenomenon. This is one contribution of the present study. In addition to that, this model also allows the determination of the structural failure mode and its change along time, which is the main contribution of this work.

2 PHYSICAL NONLINEARITY OF MATERIALS

2.1 DAMAGE MECHANICS MODEL FOR CONCRETE

The nonlinear mechanical behavior of concrete originates from the crack growth along the concrete mass [36,37]. To represent this singular mechanical behavior, damage models are particularly suitable, because this approach allows the consistent representation of material degradation by penalizing the material stiffness as a function of strain increase. In this study, the Mazars' damage model [38] was adopted, which is based on the following hypotheses: damage is an isotropic and scalar variable, the residual strains are totally neglected and damage occurs only by elongation strains.

The state of elongation at a given point is represented by the equivalent strain, as follows:

$$\tilde{\varepsilon} = \sqrt{(\varepsilon_1)_+^2 + (\varepsilon_2)_+^2 + (\varepsilon_3)_+^2} \quad (5.1)$$

in which $(\varepsilon_i)_+$ corresponds to the positive components of the main strain tensor. These components are defined by: $(\varepsilon_i)_+ = [\varepsilon_i + |\varepsilon_i|]/2$ with $(\varepsilon_i)_+ = \varepsilon_i$ if $\varepsilon_i > 0$ or $(\varepsilon_i)_+ = 0$ if $\varepsilon_i \leq 0$.

The criterion used to verify the material mechanical integrity is defined by:

$$f = \tilde{\varepsilon} - \tilde{S}(D) < 0 \quad (5.2)$$

The variable $\tilde{S}(D)$ represents the limit strain value as a function of the damage state. At the beginning of the incremental-iterative process, $\tilde{S}(D)$ corresponds to the normal strain value of concrete tensile strength, ε_{d0} . For the following steps, $\tilde{S}(D)$ is updated considering the value of $\tilde{\varepsilon}$ determined on the last converged load step, which accounts for the updated damage state. Due to the nonsymmetric mechanical behavior of concrete when subjected to tensile and compressive stresses, the damage variable is composed by the sum of two independent parts: tensile portion, D_T , and compressive portion, D_C . Each of these portions indicates tensile and compressive contributions to the local strain state. These portions are obtained as a function of the equivalent strain and the internal parameters of the damage model, which are defined by:

$$\begin{aligned} D_T &= 1 - \frac{\varepsilon_{d0}(1 - A_T)}{\tilde{\varepsilon}} - \frac{A_T}{e^{[B_T(\tilde{\varepsilon} - \varepsilon_{d0})]}} \\ D_C &= 1 - \frac{\varepsilon_{d0}(1 - A_C)}{\tilde{\varepsilon}} - \frac{A_C}{e^{[B_C(\tilde{\varepsilon} - \varepsilon_{d0})]}} \end{aligned} \quad (5.3)$$

in which $\varepsilon_{d0}, A_T, B_T, A_C, B_C$ are the internal parameters of Mazars' damage criterion. The subscripts T and C indicate tensile and compressive parts, respectively.

After the determination of each portion of damage, the final value of the damage variable is composed as follows:

$$D = \alpha_T D_T + \alpha_C D_C \quad (5.4)$$

The coefficients α_T and α_C are calculated using the following expressions:

$$\alpha_T = \frac{\sum_i (\varepsilon_{T_i})_+}{\varepsilon_V^+}, \quad \alpha_C = \frac{\sum_i (\varepsilon_{C_i})_+}{\varepsilon_V^+} \quad (5.5)$$

in which ε_{T_i} and ε_{C_i} are determined from the main stresses considering elastic material and ε_V^+ represents the total state of elongation, which is equal to:

$$\varepsilon_V^+ = \sum_i (\varepsilon_{T_i})_+ + \sum_i (\varepsilon_{C_i})_+ \quad (5.6)$$

After the damage variable determination, the stress state for the structure (solid) is calculated as follows:

$$\begin{aligned} \sigma &= (1 - D)E\varepsilon \\ \tau &= (1 - D)G\gamma \end{aligned} \quad (5.7)$$

in which E and G are longitudinal and transversal elasticity modules, respectively, and ε and γ are normal and distortional strains, respectively.

2.2 PLASTICITY MODEL FOR STEEL

Steel, as most part of ductile materials, presents elastic mechanical behavior until reaching its yield stress. After this stress level, there are some movements in the internal crystals of the material, which give it a new strength capacity. In this phase, named hardening, there is loss of stiffness, but the material still has strength capacity until its failure limit. The models based on elastoplasticity theory are appropriate to describe such mechanical behavior. Thus, the model chosen to describe the mechanical behavior of reinforcements' steel is composed by an elastoplastic constitutive law with positive isotropic hardening. The criterion used to verify the elastoplastic steel behavior is given by:

$$f = \sigma_s - (\sigma_{sy} + K\psi) \quad (5.8)$$

in which σ_s indicates the steel reinforcements' layer stress, σ_{sy} is the steel yielding stress, K is the hardening plastic modulus, and ψ is the equivalent plastic strain measurement.

The stress over each reinforcement layer is determined as follows:

$$\begin{aligned} f \leq 0 &\rightarrow \sigma = E\varepsilon \\ f > 0 &\rightarrow \sigma = E_t\varepsilon \end{aligned} \quad (5.9)$$

in which E_t is the tangent elastoplastic modulus, which is given by: $E_t = EK / (E + K)$.

3 GEOMETRIC NONLINEARITY

Figure 5.2 illustrates the initial and final configurations for a given point P, which belong to a solid, after loading action. The horizontal and vertical displacements of such point are defined by:

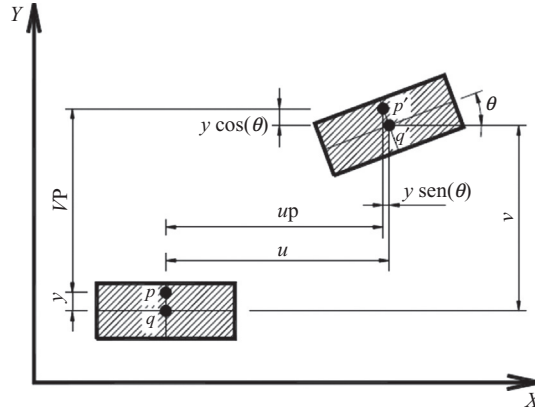
$$\begin{aligned} u_p(x, y) &= u(x) - y \sin(\theta) \\ v_p(x, y) &= v(x) - y \cos(\theta) \end{aligned} \quad (5.10)$$

Considering a second-order approximation for displacements, where $\sin(\theta) = v'(x)$ and $\cos(\theta) = 1 - \frac{v'^2(x)}{2}$, Equation 5.10 is rewritten as follows:

$$\begin{aligned} u_p(x, y) &= u(x) - yv'(x) \\ v_p(x, y) &= v(x) - y\left(\frac{v'(x)^2}{2}\right) \end{aligned} \quad (5.11)$$

in which u and v correspond, respectively, to horizontal and vertical displacement fields for any point along the body.

Considering that geometric nonlinearity second-order terms are given by Green strain measurement, the normal and distortional strain fields, ε_{xx} and γ_{xy} , respectively, are written as follows:

**FIGURE 5.2**

Initial and final configuration of a given point.

$$\begin{aligned}\epsilon_{xx} &= \frac{\partial u_p}{\partial x} + \frac{1}{2} \left[\left(\frac{\partial u_p}{\partial x} \right)^2 + \left(\frac{\partial v_p}{\partial x} \right)^2 \right] \\ \gamma_{xy} &= \frac{\partial u_p}{\partial y} + \frac{\partial v_p}{\partial x} + \left(\frac{\partial u_p}{\partial x} \frac{\partial u_p}{\partial y} + \frac{\partial v_p}{\partial x} \frac{\partial v_p}{\partial y} \right)\end{aligned}\quad (5.12)$$

Equations 5.11 and 5.12 provide the final expression for strains field, which is written as a function of displacements for the frame finite element as follows:

$$\begin{aligned}\epsilon_{xx} &= u' + \frac{1}{2} (u')^2 + \frac{1}{2} (v')^2 - yv''(1 + u') \\ \gamma_{xy} &= v' - \varphi - u'v' - \frac{v'^3}{2}\end{aligned}\quad (5.13)$$

in which φ is the additional rotation term due to Timoshenko's kinematics.

Green's strain tensor is naturally conjugated by the second Piola-Kirchhoff stress tensor. However, in the context of small displacements and strains, the second Piola-Kirchhoff stress tensor is replaced by the conventional stress tensor. Then:

$$\underline{\underline{S}} = \underline{\underline{D}}_0 \left\{ \begin{matrix} \epsilon_{xx} \\ \gamma_{xy} \end{matrix} \right\} \quad (5.14)$$

in which $\underline{\underline{S}}$ is the conventional stress tensor with normal and shear components and

$\underline{\underline{D}}_0$ is the material elastic properties tensor written as $\underline{\underline{D}}_0 = \begin{bmatrix} E & 0 \\ 0 & G \end{bmatrix}$.

The updated Lagrangian formulation describes the structure position based on the last equilibrium configuration. Thus, all information required for the next load step is obtained from the last converged load step. The numerical procedure concerns two updates: positions of each structural node and stresses at each integration point along the finite element length. The stress tensor is updated by relating Cauchy's tensor with the second Piola-Kirchhoff stress tensor. However, for small displacements

and strains, Cauchy's tensor in the current configuration coincides with the second Piola-Kirchhoff tensor of the last equilibrium configuration. Thus, the update process for stress occurs by adding the extra stress of the current step to the last step stress values, as follows:

$$\begin{aligned} x &= x_a + \Delta x \\ y &= y_a + \Delta y \end{aligned} \quad (5.15)$$

$$\begin{aligned} \sigma_{xx} &= \sigma_{xxa} + \Delta\sigma_{xx} \\ \tau_{xy} &= \tau_{xya} + \Delta\tau_{xy} \end{aligned} \quad (5.16)$$

in which x_a and y_a are nodes positions along x and y directions of the last step, Δx and Δy are displacements calculated on the current step, σ_{xxa} and τ_{xya} are normal and shear stresses of the last step, and $\Delta\sigma_{xx}$ and $\Delta\tau_{xy}$ extra stresses determined in the current step.

4 SHEAR STRENGTH MODEL

There is still a lack of knowledge in the modeling of shear effects based on finite beam elements applied to the representation of mechanical behavior of reinforced concrete structures. Therefore, the improvements presented in this item compose one contribution of this study.

Figure 5.3 illustrates a cracked reinforced concrete member and the shear transfer mechanisms considered by the proposed model. The mechanical shear resistance is composed by concrete and reinforcements contributions. The concrete shear-resistant component, V_C , is composed by V_i and V_a parts, which are related to intact concrete and interlock aggregate phenomenon, respectively. The resistant shear component due to longitudinal and transversal reinforcements is composed by dowel action phenomenon, V_d , and pure shear reinforcements parts, V_{sw} , respectively.

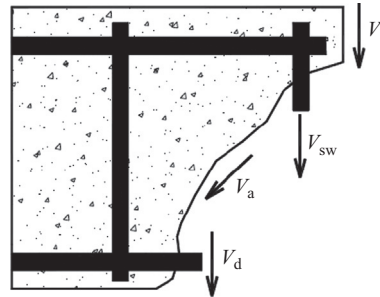


FIGURE 5.3

Cracked reinforced concrete member and shear transfer mechanisms.

4.1 INTACT CONCRETE AND AGGREGATE INTERLOCK RESISTANT CONTRIBUTIONS

The shear-resistant contributions due to concrete are given according to the following criterion:

$$\begin{aligned} D = 0 &\rightarrow V_C = V_i \\ 0 < D < 1 &\rightarrow V_C = V_a \end{aligned} \quad (5.17)$$

One of most used approaches to account for aggregate interlock is by reducing the transversal elasticity modulus by a factor that depends, essentially, on the diagonal opening cracks [39,40]. The opening crack measurement is approximated using the main tensile strain value, ε_1 . Therefore, the updated value of transversal elasticity modulus is calculated assuming a penalization parameter, μ , for which the value is in between 0 and 1. This parameter depends on the value of ε_1 and allows updating G as μG .

As the damage variable at a given point is a function of its main strain state, as previously presented, strain ε_1 also governs directly the reduction on concrete transversal stiffness. Thus, intact concrete and aggregate interlock strength contributions are assessed by the integration of shear stresses along the cross-section of reinforced concrete finite elements as follows:

$$\begin{aligned} V_i &= \int_{-h/2}^{h/2} G\gamma_{xy} dy \\ V_a &= \int_{-h/2}^{h/2} (1-D)G\gamma_{xy} dy \end{aligned} \quad (5.18)$$

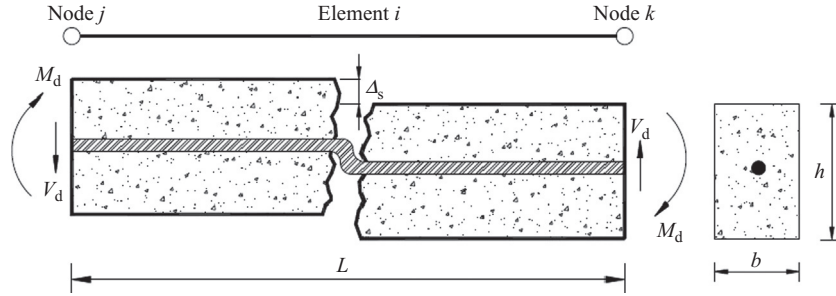
in which h is the cross-section height.

Therefore, by using convenient Mazars' damage parameters, A_T, B_T, A_C and B_C , based on experimental analysis, for instance, aggregate interlock is automatically represented as the transversal mechanical stiffness reduction is consistently modeled using damage D .

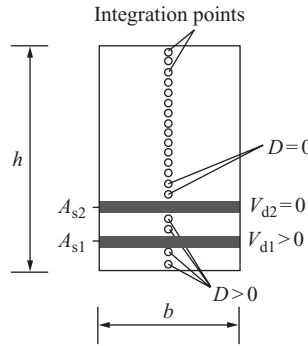
4.2 DOWEL ACTION STRENGTH CONTRIBUTION

The dowel action refers to a shear strength complementary mechanism attributed to reinforced cracked concrete. This shear-load transfer mechanism occurs when cracks grow and cut across longitudinal reinforcements, providing an increase into the mechanical shear strength. Then, the crack lips transfer shear stresses to reinforcements. As a result, a local bending and shear at reinforcements are observed. The dowel action can increase significantly the structural shear strength as well as the post-peak ductility of some structural elements, such as beams with few or no shear reinforcements.

In the proposed nonlinear model, reinforcements are modeled as beams over an elastic foundation composed by concrete. Therefore, the dowel action behavior may be affected by several factors, such as reinforcements position along the

**FIGURE 5.4**

Dowel action mechanism along a reinforced cracked concrete member.

**FIGURE 5.5**

Criterion for dowel action along the cross-section.

cross-section, concrete cover thickness, and longitudinal and transversal reinforcements ratio. Figure 5.4 illustrates the development of dowel forces into a reinforced cracked concrete member.

The bending moment caused by dowel action is determined as follows:

$$M_d = V_d L \quad (5.19)$$

in which V_d is the dowel shear force and L is the finite element length.

The criterion that defines the start of dowel action contribution is given by Mazars' damage approach. Figure 5.5 presents the application of this damage criterion in the modeling of dowel action contribution. The presence of mechanical damage is verified at integration points immediately before and after each reinforcement layer. If both points are damaged, that reinforcement layer contributes with dowel action force.

He and Kwan [39] proposed an interesting approach to estimate the dowel force, V_d . According to this work, the dowel force is calculated as follows:

$$V_d = E_s I_s \lambda^3 \Delta_s \leq V_{du} \quad (5.20)$$

The parameters involved in the evaluation of Equation 5.20 are given by:

$$I_s = \frac{\pi \varphi_s^4}{64}, \quad \lambda = \sqrt[4]{\frac{k_c \varphi_s}{4 E_s I_s}}, \quad k_c = \frac{127 c \sqrt{f_c}}{\sqrt[3]{\varphi_s^2}} \quad (5.21)$$

in which E_s is the longitudinal elasticity modulus of steel, I_s is the inertia moment of a circular cross-section, φ_s is the reinforcement diameter, λ is a parameter that compares the stiffness of surrounded concrete with reinforcement stiffness, Δ_s is the dowel displacement, k_c represents the stiffness coefficient of the surrounding concrete, f_c is the concrete compressive strength and c is an experimental parameter that reflects the spacing between reinforcements. Values between 0.6 and 1.0 may be assumed, according to He and Kwan [39].

The dowel strength is limited by the reinforcement shear resistance, which is determined as follows:

$$V_{du} = 1.27 \varphi_s^2 \sqrt{f_c} \sqrt{\sigma_{sy}} \quad (5.22)$$

in which σ_{sy} represents the yield stress of longitudinal reinforcements.

The dowel displacement of a finite element is determined by averaging the strain values assessed for each integration point, as proposed by Enright and Frangopol [27]. Then:

$$\Delta_s = \frac{\sum_{i=1}^{n_{ht}} \left\{ \frac{\pi}{\lambda} [\varepsilon_1 \cos(\alpha) \sin(\alpha) + \gamma_{xy} \cos^2(\alpha)] \right\}_i}{n_{ht}} \quad (5.23)$$

in which α is the main tensile direction defined over the horizontal plane and n_{ht} is the number of integration points along the cross-section of a finite element.

4.3 SHEAR REINFORCEMENTS STRENGTH CONTRIBUTION

Traditional structural modeling based on one-dimensional finite beam elements does not account the mechanical strength contribution of transversal reinforcements, named as stirrups. In this regard, to improve the mechanical representation provided by the numerical approach, it is necessary to introduce into the proposed nonlinear FEM formulation the presence of transversal reinforcements.

For bended finite beam elements with high span-to-depth ratio, the bidimensional stress state causes an increase on material damage, which is assessed based on shear and normal stresses intensities. In such cases, the concrete lost faster its mechanical stiffness. Therefore, the presence of shear reinforcements is required to guarantee the resistant capacity of cross-sections, especially on shear solicitations.

According to Belarbi and Hsu [41], transversal reinforcements are subjected to significant strains only after the beginning of diagonal concrete cracking. Before of such cracking, concrete resistance is composed by intact concrete, over the non-damaged region, and aggregate interlock mechanism, over the damaged regions.

The diagonal cracks opening into concrete is directly associated with the main tensile strain, ε_1 . In the same way, Mazars' damage criterion is calculated based on the tensile part of main strain tensor, which allows assuming that stirrups will be loaded, effectively, only after the beginning of concrete damage.

Therefore, the criterion that defines the beginning of shear reinforcement mechanical contribution along the load process is given by the damage criterion expressed in Equation 5.2. According to this approach, a part of shear force dissipated by damage effect is transferred to stirrups, as shown in Figure 5.6. While the equivalent strain does not reach the limit imposed by the damage criterion, the shear force on stirrups is null. After reaching this limit, the total strain is divided into two parts as follows:

$$\varepsilon = \varepsilon_e + \varepsilon_d \quad (5.24)$$

in which e represents the elastic strain portion and d is the dissipated portion.

Based on the content of Equation 5.7, one writes the dissipated strain portion as $\varepsilon_d = D\varepsilon$. Similarly, the damaged stress portion is defined as $\sigma_d = DE\varepsilon$.

The Ritter-Mörsch's truss model was used to evaluate the force part transferred to stirrups. Sanches and Venturini [42] assumed the stress state at the middle of stirrup to define its strain state. However, when nonlinear mechanical behavior is assumed, the stresses on stirrups increase from compressed flange toward the tensile flange. Moreover, it decreases in regions near to longitudinal reinforcements. Points located between the cross-section central line and the closest reinforcement layer must be verified because their strains may be larger than those obtained in the cross-section middle point.

Therefore, to better describe the equilibrium strain condition mentioned earlier, the cross-section point with the largest strain was adopted. Moreover, it was assessed

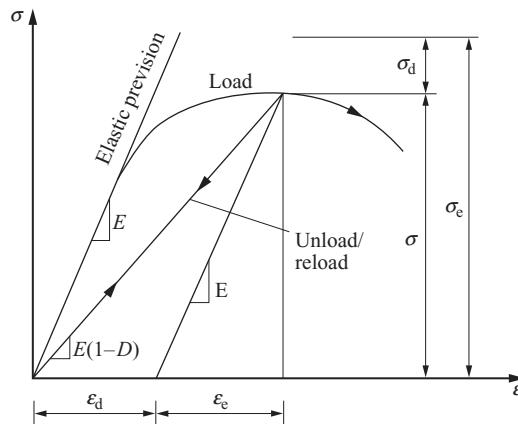


FIGURE 5.6

Scheme for stress transfer from concrete to stirrups.

by the maximum value of the rotated main strain damaged portion toward the reinforcement direction. Mathematically, this condition is expressed as follows:

$$\varepsilon_{sw} = \max[\varepsilon_1 D \sin(\alpha)] \quad (5.25)$$

in which ε_{sw} represents the stirrup strain, α is the main tensile direction, and $\max$ indicates the maximum operator.

The resultant shear force on each stirrup is assessed by $\sigma_{sw}A_{sw}$, where A_{sw} corresponds to a single stirrup cross-section area and σ_{sw} the stirrup normal stress. σ_{sw} value is obtained using the elastoplastic model over total stirrup strain, ε_{sw} . According to the Ritter-Mörsch's truss model, the shear force resisted by stirrups is calculated for a range of width equal to the effective depth of section, d . Therefore, the shear reinforcement contribution is written as follows:

$$V_{sw} = \sigma_{sw}\rho_{sw}bd \quad (5.26)$$

in which ρ_{sw} is the transversal reinforcement ratio defined by $A_{sw}/(sb)$, s is the spacing between stirrups, and b is the cross-section width.

4.4 SHEAR RESISTANCE IMPROVEMENTS INTO THE NONLINEAR SOLUTION TECHNIQUE

The numerical formulation proposed in this study is nonlinear since displacements, internal efforts, and external forces have a nonlinear dependency due to the representation of the mechanical degradation as a function of the loading process. This nonlinear problem is solved using the Newton-Raphson technique, which involves prevision and correction steps. The stiffness matrix is evaluated considering the actual damage state at each integration point and the loading process is transformed into an incremental-iterative procedure. The nonequibrated force vector is obtained by the difference between the forces due to the applied load and the resistant forces, which reflect the admissible equilibrium state considering damage and elastoplasticity criteria.

The shear forces due to the external load are mechanically equilibrated by concrete and reinforcements strength components, as previously presented. The total shear mechanical resistance is provided by concrete, either intact or aggregate interlock, reinforcements through dowel action, and reinforcements resistance itself. Therefore, the shear mechanical resistance is expressed as follows:

$$V = V_c + V_d + V_{sw} \quad (5.27)$$

The improvements proposed on the representation of shear strength mechanisms reflect the admissible shear force on the equilibrium configuration. As a result, all improvements proposed are accounted during the incremental-iterative nonlinear process and during the determination of the nonequibrated force vector.

The stiffness matrix is updated considering the actual mechanical damage state of concrete and reinforcements. Therefore, the update process concerns the values of longitudinal elasticity modulus of concrete and elastoplastic modulus of

reinforcements. It is worth to mention that all influence of the improvements on shear representation is accounted during the nonlinear procedure.

5 FICK'S DIFFUSION LAW

Corrosion of reinforcements induced by chlorides occurs in the presence of oxygen and moisture when the chloride build-up within the structures exceeds a threshold value. Even for carefully constructed concrete structures, with negligible or practically non-chloride inherited at the construction stage, the gradual build-up of chloride content takes place slowly through ingress of chlorides from external sources.

The transport phenomenon associated with the movement of chlorides along structures exposed to aggressive environments is attributed, in most part, to diffusion of chloride ions into concrete pores under a concentration gradient. The coefficient of chloride diffusion, which depends upon the pore structure of concrete, characterizes such flow under a given external concentration of chloride. This parameter is assumed as a characteristic property of hardened concrete.

To simulate the chloride ingress and its transport into concrete pores, Fick's diffusion law [43] has been widely considered as an acceptable model [2,9,15,17,20,32,34,35]. Fick's laws for diffusion are applicable for homogeneous, isotropic, and inert materials [44]. Moreover, the mechanical properties related to diffusion process are assumed to be identical along all directions and kept constants along time. Considering concrete, these hypotheses are not completely satisfied, because concrete is well known as heterogeneous, anisotropic, and chemically reactive (continued hydration and microcracking process) material. However, the methods commonly adopted for chlorides transportation modeling in concrete consider this process governed only by ionic diffusion. Then, it assumes that the concrete cover is completely saturated. Therefore, it makes the hypotheses of Fick's law acceptable for the chloride ingress modeling, because, in this case, the material is assumed completely saturated, with unidirectional chloride flux, that is, from the exterior surface into the concrete depth. When chloride diffuses into concrete, a change in chloride concentration, C , occurs at any time, t , in every point, x , of the concrete, that is, it is a non-steady state of diffusion. To simplify its analysis, the diffusion problem is considered as one-dimensional. Many engineering problems of chloride ingress, as those discussed in this study, can be solved considering this simplification.

Fick's diffusion theory assumes that the transport of chlorides into concrete through a unit section area of concrete per unit of time (the flux F) is proportional to the concentration gradient of chlorides measured at normal direction of section. Then:

$$F = -D_c \frac{\partial C}{\partial x} \quad (5.28)$$

The negative sign on the equation above arises because the diffusion of chlorides occurs in the opposite direction of the increasing concentration of chlorides. The

constant of proportionality D_c , presented in Equation 5.28, is called chloride diffusion coefficient. In general, D_c is not constant, but depends on many parameters as the time for which diffusion has taken place, location in the concrete, composition of concrete, among other factors. If the chloride diffusion coefficient is constant, Equation 5.28 is usually referred as Fick's first diffusion law. If this is not the case, the relation is referred as Fick's first general diffusion law.

There are some cases where this simple relation should not be applied. In this regard, it is worth to mention the cases where the diffusion process may be irreversible or has a history dependence. In such cases, Fick's diffusion law is not valid and the diffusion process is referred as anomalous. However, nonobservation so far indicates that the chloride diffusion into concrete pores should be characterized as an anomalous diffusion. Fick's second law can be derived considering the mass balance principle. Therefore,

$$\frac{\partial C}{\partial t} = \frac{\partial}{\partial x} \left(-D_c \frac{\partial C}{\partial x} \right) \quad (5.29)$$

To apply Fick's second diffusion law, in this form, for concrete exposed to chloride during a long period of time, one ought to know the variation of the chloride diffusion coefficient along time. If only few observations exist in a specific case, it is possible to estimate upper and lower boundaries for the variation of D in time. In spite of this dependence, an especial case can be considered where the chloride diffusion coefficient is independent of location, x , time, t , and chloride concentration, C . In this case, Fick's second law is written in this simple form:

$$\frac{\partial C}{\partial t} = D_0 \frac{\partial^2 C}{\partial x^2} \quad (5.30)$$

in which D_0 is the constant coefficient of diffusion.

The solution of the differential equation presented above, for a semi-infinite domain with a uniform concentration at the structural surface, is given as follows:

$$C(x, t) = C_0 \operatorname{erfc} \left[\frac{x}{2\sqrt{D_0 t}} \right] \quad (5.31)$$

in which C_0 is the chloride concentration at the structural surface supposed to be constant in the time; erfc is the complementary error function.

In a physical sense, field conditions deviate significantly from the assumptions implicit in Fick's law. For instance, the cover is not always saturated with water, and so chloride ions penetrate concrete by diffusion and advection provided by the penetrating moisture front. Concrete is not homogeneous due to the presence of microcracking and interconnected pores, then, the diffusion coefficient will change with time as hydration proceeds. Hence, Fick's law is not an excellent model for this phenomenon. Nonetheless, Fick's law is often used since in many cases the diffusion equation provides the best approximation to laboratory or field data [45]. Clearly, predictions using such approach are valid only if best-fit parameter values are applied to structures with similar material, environmental, and field conditions. It is preferable, in using this approach that concentrations are given in terms

of water-soluble chlorides since it is generally accepted that corrosion is influenced by the free chloride concentration present in the concrete pore solution [46].

As one-dimensional finite beam element is adopted in this study, it seems to be consistent in representing the chloride ingress in the same dimension, that is, in one-dimensional sense. In this work, Equation 5.31 is used to evaluate the time in which the corrosion process starts. After the corrosion time initiation, the reinforcements' areas are penalized using the relations that are presented in next section.

6 CORROSION LAWS

The classical corrosion model proposed by Tuutti [46] states that reinforcements corrosion occurs into two distinct stages: initiation period and propagation period. The initiation period is limited by the time required for the occurrence of reinforcements depassivation due to the chloride concentration growth along time. The propagation period comprises the time in which cross-section area of reinforcements steel is lost. Figure 5.7 illustrates the evolution of cross-section diameter of reinforcement, D_s , as a function of time, where initiation and propagation periods are clearly defined.

The rate of reinforcements corrosion into propagation period depends on environmental moisture and the amount of available oxygen. These two parameters have major importance on the cathodic process and on the concrete resistivity. In addition to them, temperature catalyzes the corrosion process.

In spite of its importance, mathematical approaches for modeling the propagation period of corrosion are not often available in literature, because many researchers consider the initiation period as structural life time. However, the reduction of steel is consistently represented by models presented in Refs. [17,34,35]. The models presented on mentioned references were determined empirically, accounting for tropical climate.

In the present study, the steel loss is evaluated using the models presented in Refs. [17,34,35], which assumes that steel is corroded either uniformly or by pits along its perimeter. For uniform corrosion model, the corroded diameter of reinforcements is determined as follows:

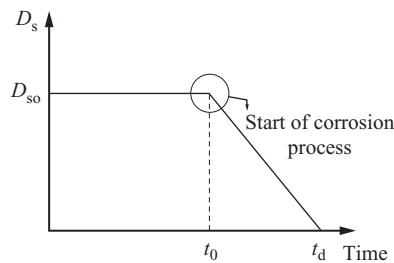


FIGURE 5.7

Change on reinforcements diameter along time considering corrosion effects.

$$d(t) = \begin{cases} d_{\text{initial}} & \text{if } t \leq t_0 \\ d_{\text{initial}} - 0.0232i_{\text{CORR}}(t - t_0) & \text{if } t > t_0 \end{cases} \quad (5.32)$$

in which d_{initial} is the reinforcement diameter before depassivation in mm, t_0 represents the time of corrosion initiation in years, and i_{CORR} indicates the corrosion ratio given by $\mu\text{A}/\text{cm}^2$.

The corrosion rate is calculated using the expression presented in Ref. [35], which was also determined empirically.

$$i_{\text{CORR}} = \frac{37.8(1 - w/c)^{-1.64}}{\text{cvt}} (\mu\text{A}/\text{cm}^2) \quad (5.33)$$

in which w/c indicates the water/cement ratio and cvt represents the concrete cover thickness in cm.

On the other hand, when pitting corrosion is considered, nonuniform corrosion model is assumed. The corroded diameter of reinforcements is determined using the following expression:

$$d(t) = \begin{cases} d_{\text{initial}} & \text{if } t \leq t_0 \\ d_{\text{initial}} - 0.0116Ri_{\text{CORR}}(t - t_0) & \text{if } t > t_0 \end{cases} \quad (5.34)$$

where R represents the ratio between steel corroded thickness calculated using uniform and pitting corrosion approaches. The value 5.65 may be adopted for R variable according to Vu and Stewart [35]. It is important to mention that Equations 5.32 and 5.34 were determined considering experimental results. Therefore, such equations are suitable to inherent errors associated with equipment of measure and/or human errors. However, in spite of their inherent limitations, models better than those are not available in literature.

Reinforcements corrosion induces longitudinal cracking due to the expansive products of corrosion, which leads to spalling and/or delamination phenomena. In such case, experimental results suggest that structural mechanical capacity is reduced due to the concrete compression zone being reduced by the depth of the spalled/delaminated cover. In this regard, it is important to mention that the equations presented in this section accounts only the steel loss itself, along time, after the time of corrosion initiation. Complementary mechanical degradation effects due to corrosion such as reduction of steel strength and ductility along time, spalling and delamination of concrete, and adherence loss among concrete and reinforcements are not considered by the equations adopted in this study.

7 APPLICATIONS

Two applications are considered in this study. The first application concerns the mechanical analysis of a two-dimensional frame where the results achieved by the nonlinear FEM model are compared with experimental results available in literature. In the second application, the proposed formulation is used to model the mechanical behavior of two simple supported beams subjected to chloride

penetration and reinforcements corrosion. The results obtained by the proposed formulation are compared with experimental and numerical responses available in literature.

7.1 TWO-DIMENSIONAL REINFORCED CONCRETE FRAME

The structure studied in this application is a reinforced concrete frame experimentally analyzed by Vecchio and Emara [47] and numerically tested by Guner [48] and La Borderie et al. [49]. The frame is analyzed by the nonlinear FEM model assuming that mechanical finite element is governed by pure Euler-Bernoulli approach (B), without shear contributions, and pure full Timoshenko approach (TSD), including shear contributions. The loads and frame geometry are presented in Figure 5.8.

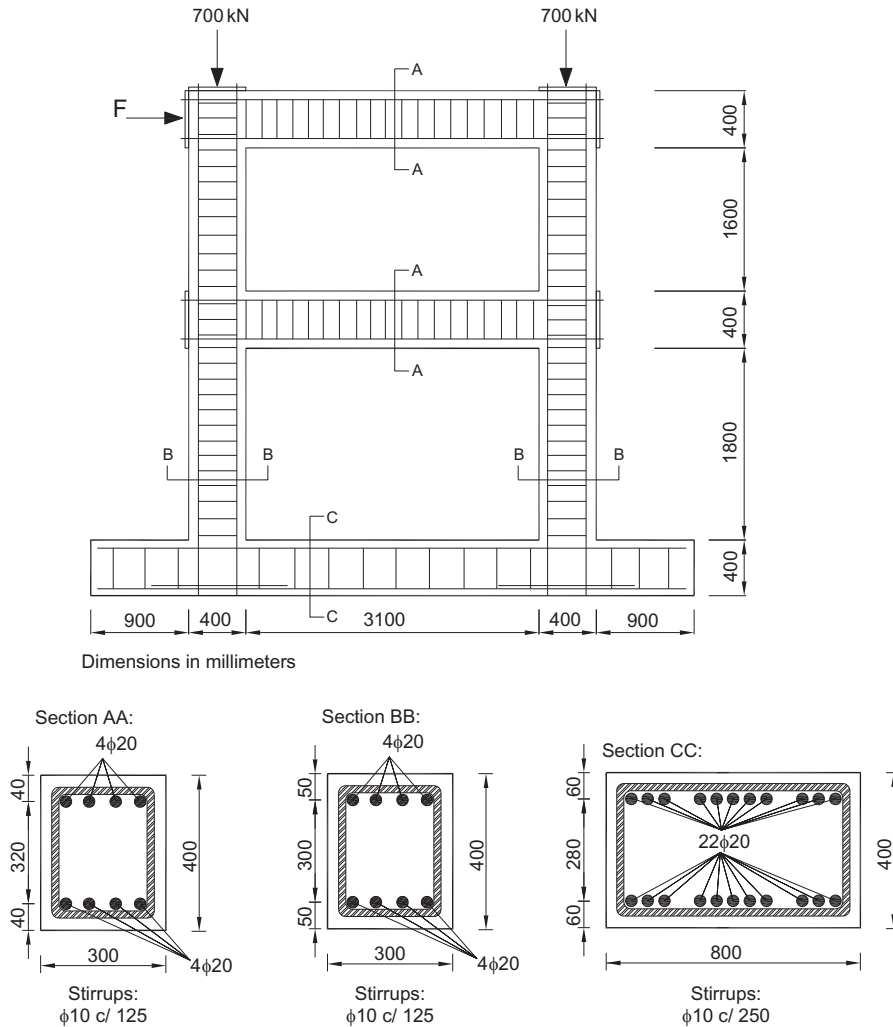
Two types of support conditions were considered: case I—frame including the support beam and case II—simple clamped-clamped frame, as illustrated in Figure 5.9. The responses determined by La Borderie et al. [49] were obtained assuming only case II. Concerning the types of analyses, Guner [48] used the software SAP 2000, in which the structure is modeled considering mixed mechanical behavior, that is, elastic-linear along the one-dimensional finite elements and plastic hinges at the appropriate member ends. Such hinges are positioned at the end nodes of some special finite elements, such as the joint of a beam and column to simulate the existence of rigid offsets. The numerical analysis performed by La Borderie et al. [49] was based on one-dimensional finite elements considering their own damage model in which the inelastic strains from the damage were accounted.

The following material parameters were adopted: Young modulus of concrete 23,674 MPa, concrete compression strength $f_{ck} = 30$ MPa, coefficient of Poisson 0.2, yield strength of steel 418 MPa, Young modulus of steel 192,500 MPa, and plastic modulus of steel 19,250 MPa. The parameters adopted for Mazars damage model were the following: $\epsilon_{d0} = 0.000085$, $A_T = 1.145$, $B_T = 10330$, $A_C = 1.117$, and $B_C = 1189$. The horizontal load on the frame top was applied into steps of 5 kN.

The equilibrium trajectories for cases I and II are presented in Figures 5.10 and 5.11, respectively. The use of clamped supports, as observed in case II, provided higher stiffness to the structure, since there was no rotation in these support nodes. The support beams adopted in case I did not introduce significant difference for model B. However, for the TSD model, considerable changes were observed in terms of displacements after concrete cracking and especially in terms of ultimate load. The high capacity of internal forces redistribution may be the main reason for this type of behavior.

Tables 5.2 and 5.3 present the values of the loading and horizontal displacement for both reinforcement steel yielding in node 1 and frame failure. The column error (%) was evaluated from a comparison between experimental and each numerical result for both yielding and ultimate loads.

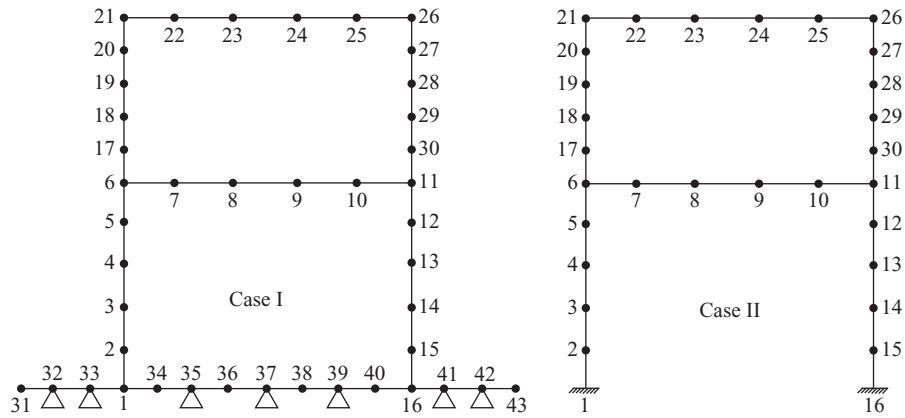
The TSD model, for case I, represented better the real structural behavior of the frame in terms of ultimate load and reinforcement steel yielding. The differences

**FIGURE 5.8**

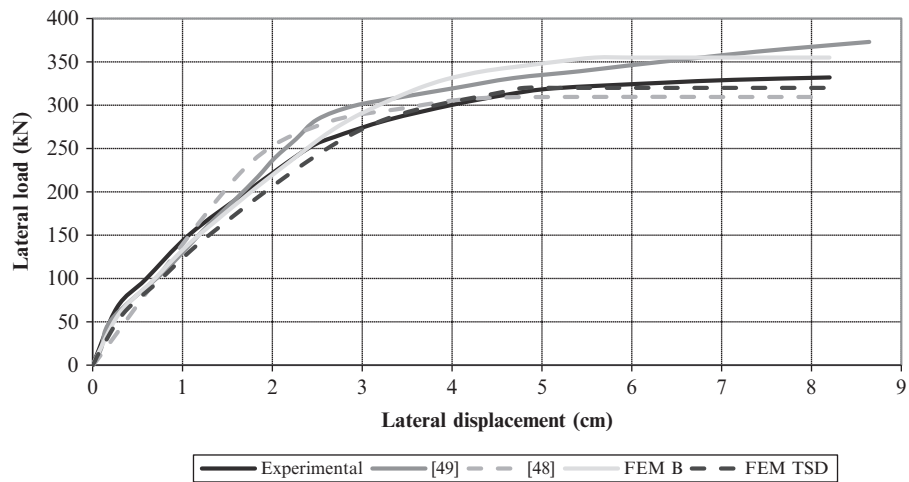
Frame analyzed.

observed are -3.6% and $+0.4\%$, respectively, in comparison to the experimental tests. In the case II, in which the structure is considered as a simple clamped-clamped frame, the proposed nonlinear model was capable to achieve good agreement to experimental results, especially for ultimate load.

Therefore, based on the results presented in this application, the proposed nonlinear FEM model was validated considering numerical and experimental results available in literature. Then, the frame considered in this application can now be analyzed including the effects of chloride penetration and reinforcements corrosion.

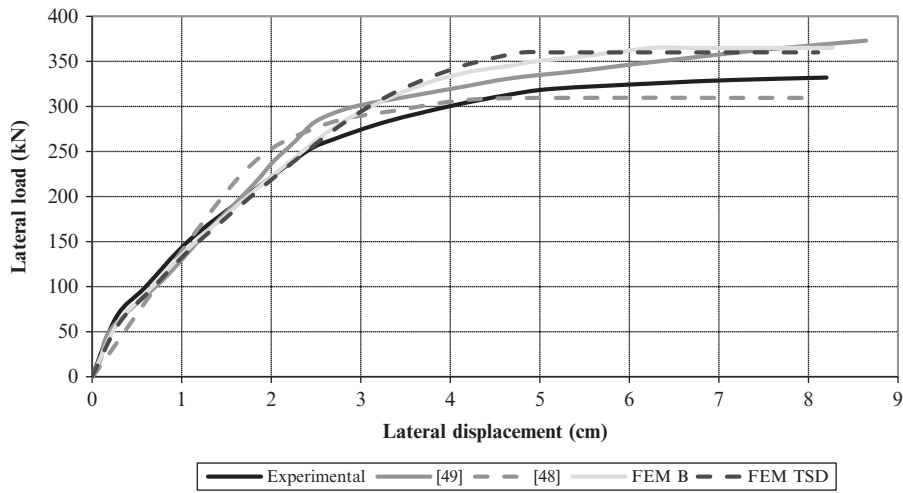
**FIGURE 5.9**

Boundary conditions case.

**FIGURE 5.10**

Horizontal equilibrium trajectory, node 21, case I.

To account for these phenomena, Fick's law was used to model the chloride penetration and chloride concentration growth along time. The penalization of rebar cross-section was modeled using uniform and pitting corrosion approaches as presented by Refs. [17,34,35]. For the analyses performed in this application, the following parameters were adopted: chloride concentration at structural surface $C_0 = 1.15 \text{ kg/m}^3$, diffusion coefficient of chlorides $D_0 = 41.0 \text{ mm}^2/\text{year}$, and threshold chloride concentration which leads to reinforcements depassivation $C(x, t) = 0.90 \text{ kg/m}^3$. Based on such values, Equation 5.31 allows calculating the

**FIGURE 5.11**

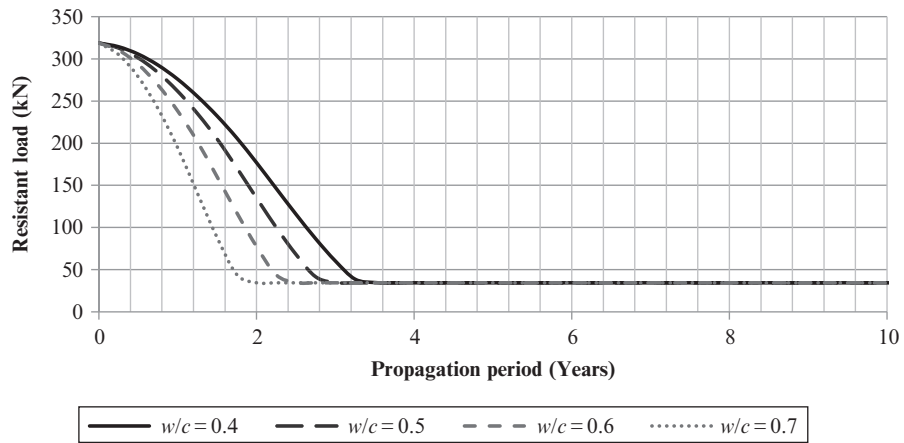
Horizontal equilibrium trajectory, node 21, case II.

Table 5.2 Comparing Results Case I

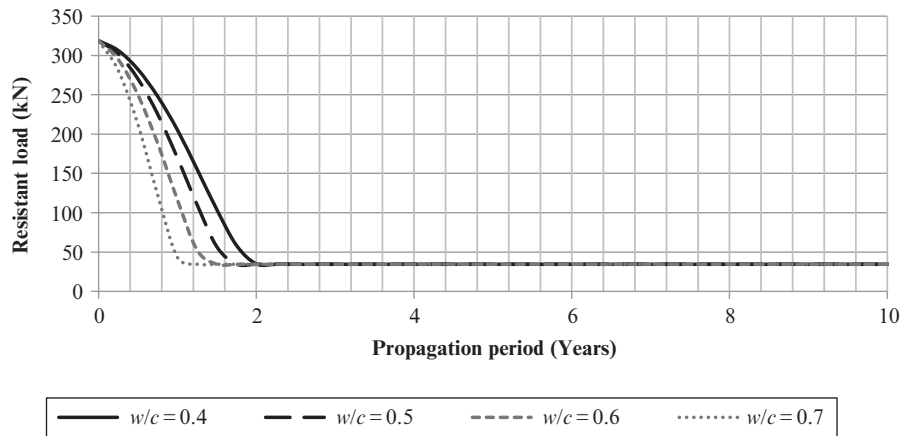
Model	Yielding			Ultimate		
	F (kN)	Displacement (cm)	Error (%)	F (kN)	Displacement (cm)	Error (%)
Experimental	264.0	2.68	0.0	332.0	8.21	0.0
Guner [48]	238.0	1.89	−9.8	309.0	8.06	−6.9
La Borderie et al. [49]	277.0	2.53	+4.9	373.0	8.64	+12.3
FEM B	285.0	2.88	+7.9	355.0	8.20	+6.9
FEM TSD	265.0	2.85	+0.4	320.0	8.31	−3.6

Table 5.3 Comparing Results Case I

Model	Yielding			Ultimate		
	F (kN)	Displacement (cm)	Error (%)	F (kN)	Displacement (cm)	Error (%)
Experimental	264.0	2.68	0.0	332.0	8.21	0.0
Guner [48]	238.0	1.89	−9.8	309.0	8.06	−6.9
La Borderie et al. [49]	277.0	2.53	+4.9	373.0	8.64	+12.3
FEM B	285.0	2.86	+7.9	365.0	8.27	+9.9
FEM TSD	283.0	2.80	+7.2	360.0	8.10	+8.4

**FIGURE 5.12**

Resistant load variation along time considering uniform corrosion.

**FIGURE 5.13**

Resistant load variation along time considering pitting corrosion.

time for corrosion initiation, which is equal to 144.16 years. The structure was analyzed considering four different water/cement ratios which are equal to 0.40, 0.50, 0.60, and 0.70.

Figures 5.12 and 5.13 illustrate the variation of the resistant structural load along time considering uniform and pitting rebar corrosion, respectively, and the four different water/cement ratios mentioned above. In these figures, only the propagation period of corrosion is considered.

As presented in Figures 5.12 and 5.13, it is observed that the water/cement ratio is directly responsible by the rate of rebar cross-section reduction along time. According to the results obtained, as bigger be the water/cement ratio smaller is the residual structural load and consequently its residual life.

In addition to that, it was also observed that the pitting corrosion approach provides more severe steel loss along time when compared to uniform corrosion model. It is worth to mention that pitting corrosion approach leads to more confident modeling when compared to experimental results as the chloride penetration and corrosion attack do not occur uniformly.

The change on the structural failure mode along propagation time is illustrated in Figures 5.14 and 5.15, considering four different water/cement ratios. As presented in these figures, the corrosion phenomenon changes the structural failure mode along time. Initially, the structural failure is governed by concrete crushing. However, as the time passes and the corrosion proceeds, the steel failure becomes the critical resistant part.

7.2 SIMPLE SUPPORTED REINFORCED CONCRETE BEAMS. CORRODED AND NONCORRODED CASES

The last application of this study concerns the mechanical analysis of two reinforced concrete beams. These beams were experimentally analyzed by Castel et al. [50] and numerically tested by Biondini and Vergani [51]. The beams scheme is illustrated in

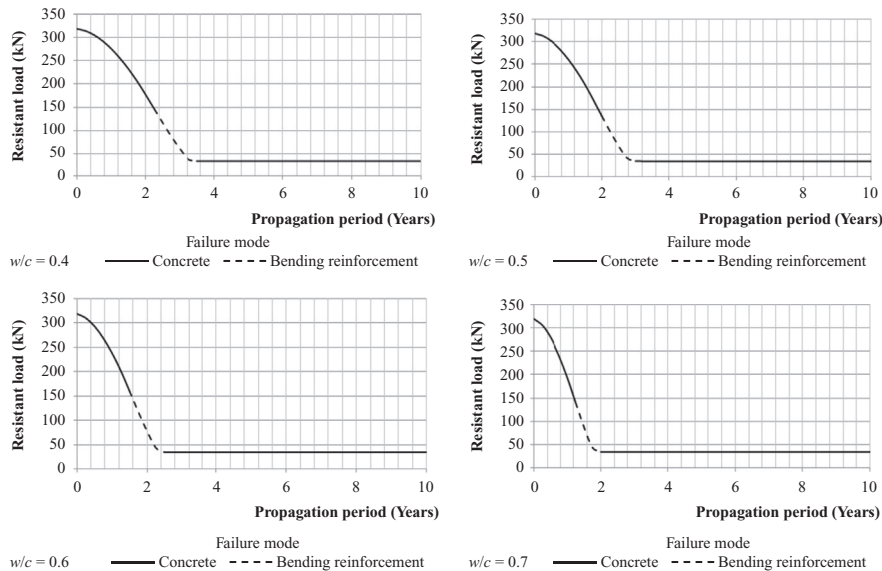
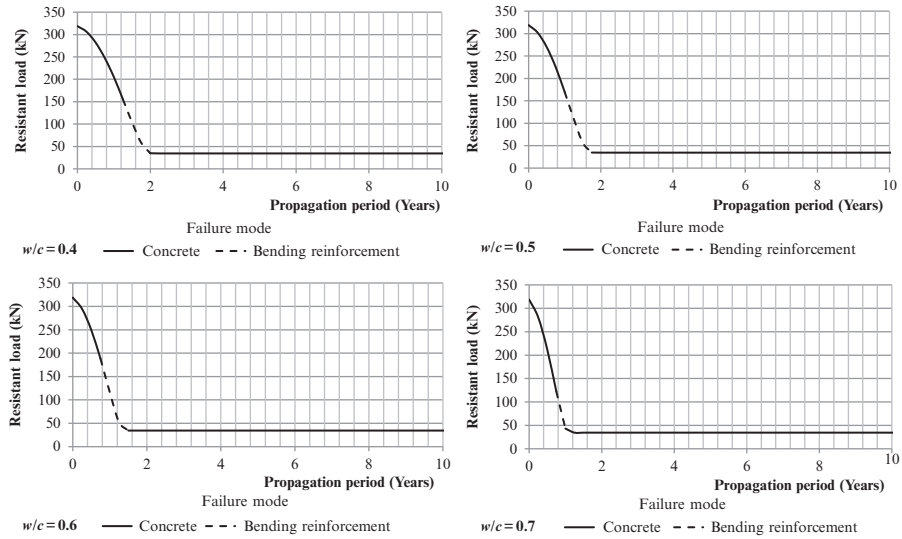
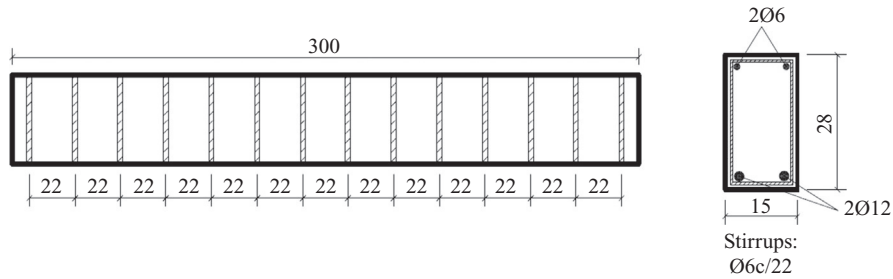


FIGURE 5.14

Failure mode change along time. Uniform corrosion approach.

**FIGURE 5.15**

Failure mode change along time. Pitting corrosion approach.

**FIGURE 5.16**

Beams scheme. Dimensions in centimeters.

Figure 5.16, which is a simple supported beam with 3.0 m span length and rectangular cross-section of 15×28 cm. The composition of concrete mixture adopted in these beams is presented by Castel et al. [50]. The two beams considered in this application are a noncorroded beam and a beam after 14 years of chloride exposure, namely, B1T and B1CL, respectively, which were modeled using the proposed nonlinear FEM model considering a concentrated load applied at mid-span and increased up to collapse. The results obtained using the nonlinear FEM model are compared with experimental, Castel et al. [50], and numerical, Biondini and Vergani [51], responses available in literature.

For this application, the following parameters were adopted: structural cover depth 10 mm and water/cement ratio 0.5, Castel et al. [50], chloride concentration at structural surface $C_0 = 35 \text{ kg/m}^3$, Castel et al. [50], diffusion coefficient of chlorides $D_0 = 63.07 \text{ mm}^2/\text{year}$, Vu and Stewart [35], threshold chloride concentration which leads to reinforcements depassivation $C(x, t) = 1.10 \text{ kg/m}^3$, Vu and Stewart [35], corrosion rate $i_{\text{corr}} = 17.81 \mu\text{A/cm}^2$, Vu and Stewart [35]. The penalization of rebar cross-section was modeled using uniform corrosion approach. Based on these values, Equation 5.31 allows calculating the time for corrosion initiation, which is equal to 0.17 years. Table 5.4 presents elasticity modulus and strength values for steel and concrete in order to simulate the mechanical behavior of both beams.

Figure 5.17 presents the comparative results for noncorroded beam, B1T, and corroded beam, B1CL. As presented in this figure, good agreement is observed

Table 5.4 Variables and Values Adopted in this Application

Beam	Elasticity Modulus of Concrete (MPa)	Concrete Strength in Compression (MPa)	Elasticity Modulus of Steel (MPa)	Yielding Steel Strength (MPa)
B1T	36,300	65.3	250,000	500
B1CL	35,000	63.44	250,000	500

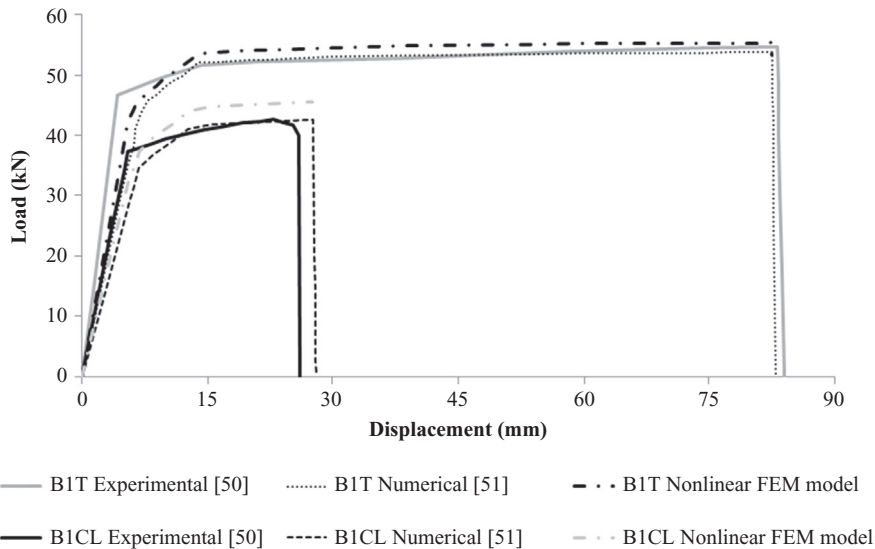


FIGURE 5.17

Comparative results for B1T and B1CL beams.

among experimental and numerical results when BIT is considered. It was expected as corrosion phenomenon is not accounted.

When BICL is analyzed, good agreement is also observed among all references considered. In this corrosion case, the proposed nonlinear model tends to slightly overestimate the structural mechanical resistance, especially, after reinforcements yielding. It may be explained due to mechanical effects such as spalling and reduction of yielding steel stress along time that are not accounted in the model.

8 CONCLUSION

In this chapter, reinforced concrete structures subjected to chloride ingress, which leads to reinforcements corrosion, were studied. The chloride ingress and its concentration growth along time were modeled using Fick's law and the corrosion of reinforcements was simulated using empirical equations presented in classical references. Uniform and pitting corrosion approaches were considered. The mechanical structural behavior was simulated using a robust and accurate nonlinear model, which is the main contribution of this chapter. The proposed nonlinear model uses algebraic FEM equations and considers the damage model of Mazars to represent the mechanical degradation of concrete. Moreover, dowel action, aggregate interlock, and stirrups stiffness are accounted. The mechanical behavior of steel is simulated using elastoplastic approach with hardening. The geometric structural nonlinearity is represented using updated Lagrangian approach.

In the first application, the results obtained by the proposed nonlinear model were compared to experimental and numerical responses available in literature. The results obtained by the proposed nonlinear model presented good agreement with those available in literature, indicating its accuracy and robustness, validating, therefore, the numerical nonlinear proposed FEM model.

Finally, the proposed nonlinear model was applied to the analysis of two reinforced concrete beams which were analyzed experimentally and numerically by other researchers, assuming corrosion and noncorrosion cases. The comparative of results indicates that the proposed nonlinear model is efficient in modeling such type of structures. However, some phenomena such as spalling and reduction of yielding steel stress along time may be considered in the future in order to improve the accuracy of the proposed model.

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Nonlinear 3D finite-element modeling for structural failure analysis of concrete sleepers/bearers at an urban turnout diamond

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1 INTRODUCTION

Railway urban turnout is a special track system used to divert a train from a particular direction or a particular track onto other directions or other tracks. It is a structural grillage system that consists of steel rails, points (or called “switches”), crossings, steel plates, rubber pads, insulators, fasteners, screw spikes, beam bearers (either timber, polymer, steel, or concrete), ballast and formation, as shown in Figure 6.1 [1]. There are two types of turnouts, a conventional turnout and a tangential turnout.

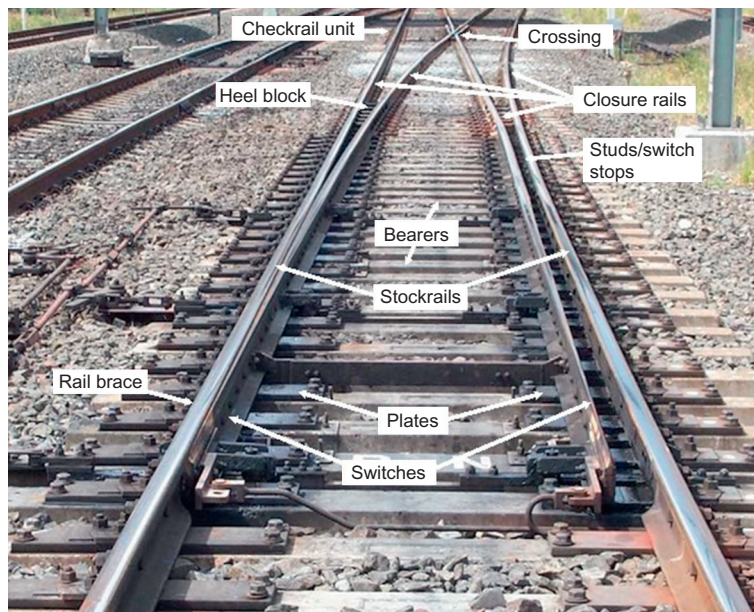


FIGURE 6.1

Components of a turnout.

Standard conventional turnouts are designed typically for straight main line track. The combination of switch length, heel angle, and cross rate defines the turnout type, and they all typically have the same components. Tangential turnouts are defined by the radius of the turnout. Components in a tangential turnout vary as manufacturers place their own designs over the standard configuration. The traditional turnout structure generally imparts high-impact forces on to its structural members because of its blunt geometry and mechanical connections between closure rails and switch rails (i.e., heel-block joints) [2].

A turnout is an inevitable structure in railway tracks whose crossing imparts a significant discontinuity in the rail running surface. The wheel/rail interaction on such imperfect contact transfer can cause detrimental impact loads on railway track and its components [1–9]. The transient vibration could also affect surrounding building structures. In addition, the large impact emits disturbing noises to railway neighbors [5,10–13]. The impact and ground-borne noises are additional to the normal rolling noise. Many previous studies have predicted impact forces and noise using numerical models [5]. However, only a few have implemented impact mitigation strategies in the field and even fewer field trial reports are available in the literature [5–13]. The impact mitigation strategies at an urban turnout include wheel/rail transverse profiling and longitudinal profiling of crossings, increased turnout resilience and damping, changes to rolling stocks, external noise/vibration controls, etc. [14–20].

Recently, there have been several incidental cracks in concrete bearers in a turnout within an urban railway network in Australia. The concrete bearer cracked under a rail pad, the part which supports the rail, where the train wheels shift over a diamond formation. The purpose of this chapter is to demonstrate an engineering failure analysis of a complex structural system using a case study of turnout diamond (Figure 6.2), which is often found in railway network. The finite-element model



FIGURE 6.2

A typical railway turnout diamond (k crossings induce double impacts on the supporting bearers).

has been developed to determine if the turnout diamond systems were sufficient for the impact loading experienced by that bearer. The emphasis of this study will be placed on the current design and analysis of turnout bearers and the application of nonlinear finite-element modeling to aid failure analysis of railway structural components.

Table 6.1 summarizes recent research on failure analysis of turnout components. It can be found that most research has paid special attention on traditional turnouts, while information on such special trackworks as diamond, single, and double slips are very limited. As a result, this study is imperative for complex turnout design, maintenance, and renewal.

2 INTERACTION OF TRAIN AND TURNOUT STRUCTURE

The wheel/rail contact over the crossing transfer zone has a dip-like shape where the wheel trajectory is not smooth. The accurate shape of the wheel trajectory (running top) and dip angle will depend on the wheel and running rail profiles. The associated dip angles, which are the acute angles between the tangents to the wheel trajectory at the point where it abruptly changes direction, can then be estimated from the wheel trajectories, as illustrated in Figure 6.3.

2.1 TURNOUT FORCES

It is generally assumed that the high-frequency impact force (P_1) that occurs either at a nose or at a wing rail has little effect on the rail foot [14]. On the other hand, the dynamic P_2 force (the second peak in the impact force history) has significant influence on the crossing components. The distance from the point of impact to the point of the peak impact force depends on a number of factors including train speed. The common damage zones to be considered are the rail foot within 0.75 m of any joints to plain rail; the base of the crossing in transfer zone (extending 0.75 m on both sides of the pick up point); and other components in vicinity of the crossings.

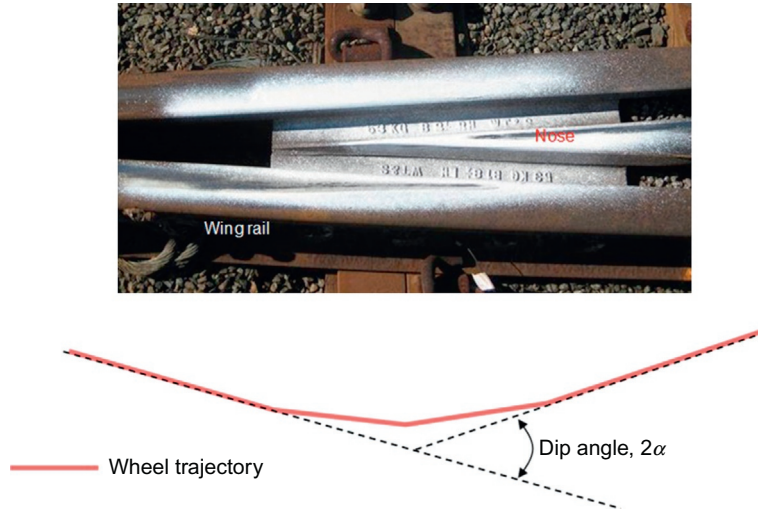
In a calculation of P_2 force, the track damping C_t is normally negligible. For plain tracks, it is commonly found that the track mass is relatively low in comparison with the wheel set mass and is then neglected. In contrast, for a turnout crossing, the track mass tends to be of significance and it cannot be neglected. Jenkins et al. [15] has proposed a formula for estimating a dynamic P_2 force as follows:

$$P_2 = P_0 + 2\alpha \cdot v \cdot \left[\frac{M_u}{M_u + M_t} \right]^{1/2} \cdot \left[1 - \frac{\pi \cdot C_t}{4\sqrt{K_t \cdot (M_u + M_t)}} \right] \cdot [K_t \cdot M_u]^{1/2}, \quad (6.1)$$

where P_2 is the dynamic vertical force (kN); P_0 is the vehicle static wheel load (kN); M_u is the vehicle unstrung mass per wheel (kg); 2α is the total joint angle or equivalent dip angle (rad); v is the vehicle velocity (m/s); K_t is the equivalent track stiffness (MN/m); C_t is the equivalent track damping (kNs/m); and M_t is the equivalent track mass (kg).

Table 6.1 Recent Research on Failure Analysis of Special Trackworks (2013–2015)

Year	Authors	Keywords	Type of Special Trackwork			
			Turnout	Crossover	Diamond	Slips
2015	Kaewunruen [8]	Bearer failure	x		x	x
2015	Tan et al. [21]	Rail switch failure	x			
2015	Pålsson and Nielsen [22]	Switch and crossing	x	x		
2014	Camci et al. [23]	Motor switch failure	x			
2014	Kaewunruen [24]	Crossing degradation	x	x	x	x
2014	Kaewunruen [25]	Composite bearer	x		x	x
2014	Ni et al. [26]	Cracked crossing	x	x		
2014	Pålsson [27]	Switch and crossing	x	x		
2014	Jönsson et al. [28]	Crossing	x	x		
2014	Jeon [29]	Formation failure	x			
2014	Wan et al. [30]	Crossing nose	x			
2014	Griffin et al. [31]	Composite track slab	x			
2014	Li et al. [32]	Turnout settlement	x	x		
2014	Chen et al. [33]	Turnout defect	x			
2014	Rapp et al. [34]	Shared corridor	x		x	
2014	Ferdous and Manalo [35]	Bearer failure	x			
2013	Pålsson [36]	Switch rail	x			
2013	Eck et al. [37]	Switch rail	x			
2013	Yang et al. [38]	Turnout parts	x			
2013	Davis and Jimenez [39]	Turnouts	x	x		

**FIGURE 6.3**

Transfer zone at crossing where a conical wheel traversing a v-crossing (removal of white paint showing the wheel contact band) and running over a dip angle inducing impact force [1,2].

2.2 TRACK STIFFNESS

Track stiffness K_t can be estimated using a beam on elastic foundation (BOEF) model. The deflection under a point load can be written as:

$$K_t = \frac{P}{y} = 8\beta^3 EI, \quad (6.2)$$

$$\beta = \left[\frac{k_f}{4EI} \right]^{1/4}, \quad (6.3)$$

where E is the Young's modulus; I is the moment of inertia of the rail or crossing section; and k_f is the foundation modulus (track modulus).

$$k_f = 4.104P_0(\text{axle load in tons}) + 14.61 \text{ MPa}. \quad (6.4)$$

2.3 EQUIVALENT TRACK MASS

Considering the bearer spacing of 600 mm, a bearer mass of 600 kg (half can be associated with the crossing), a mass per unit of the crossing is 120 kg/m, the track mass per unit rail length (m) can be estimated as 620 kg/m. The equivalent track mass (lumped) at crossing $M_t = 3m/2\beta$. Thus, M_t is varied from 705 to 756 kg.

3 FAILURE MECHANISMS

In recent years, there have been several incidents of cracking in concrete bearers in turnouts within Australian railway networks. Such problem was in relation to significant degradation of turnout components due to large impact force such as that from the Clyde 700 diamond (RailCorp's western main rail network). White ballast (ballast pulverization) could be observed near the crossings. The failure mechanism is illustrated in [Figure 6.4](#).

3.1 STRUCTURAL INTEGRITY

Initially, the early warning sign was the deviated geometry profile of the diamond, as seen in [Figure 6.4a](#). White ballast (ballast pulverization) could then be observed near v-crossings and k-crossing, as shown in [Figure 6.4b and c](#). It is believed that the ballast was pulverized due to the impact vibration of crossing structures and the bearers excited by the rail dip-like irregularity. It was also found that due to the large vibration, the wheelsets tended to spread the lubricant (as the wheel collected from friction management trackside device) over the location slightly further from the nearby crossing (such as in the crossing panel as shown in [Figure 6.4b](#)).

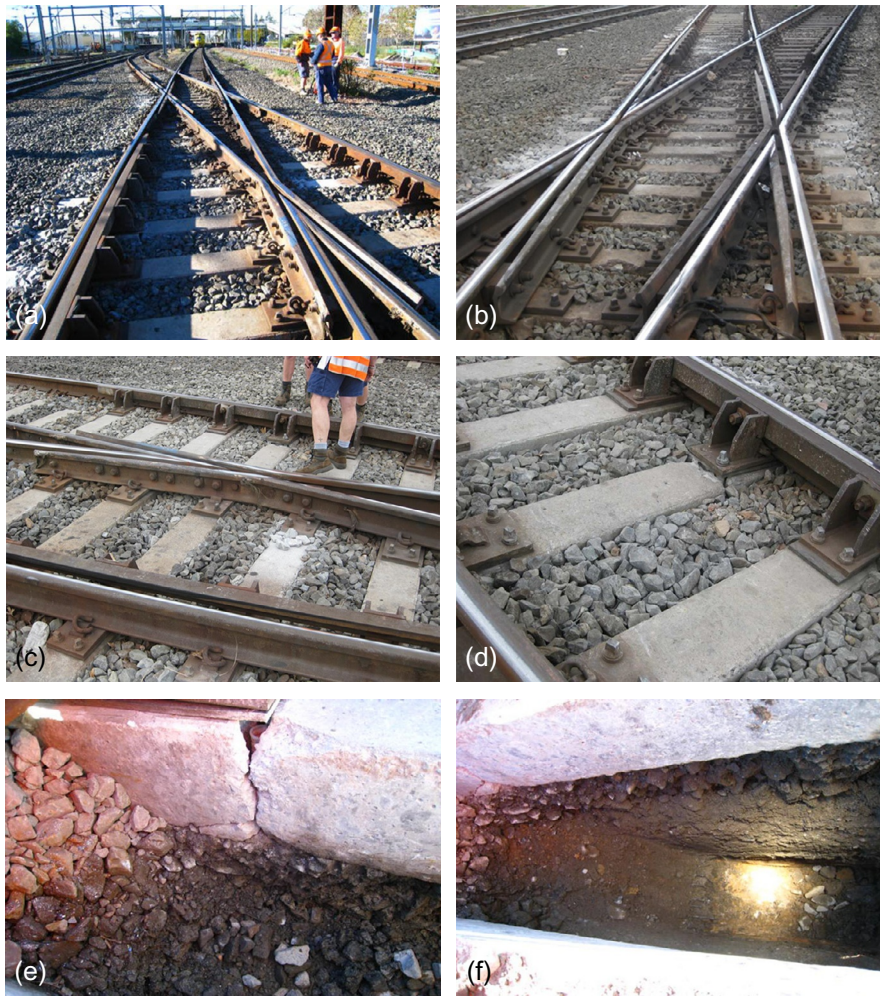
[Figure 6.4d](#) also shows the collapse of the bearer underneath the stockrail at Country-end v-crossing. This negative (hogging) bending failure is due to a single wheel impact at v-crossing. The bearer failure had led to the discrepancy of track surface (top) between running rails and the wide track gauge irregularity. The shiny steel at wing rail could be noticed due to the contact of wheelback and wing rails. Similar mechanism can also be observed at k-crossing area.

It is noted that the concrete bearer cracked under a rail pad, the part which supports the rail, where the train wheels travel over the diamond section. The concrete bearers experienced a different type of dynamic load, compared to standard railway sleepers. This is due to their inherent stiffness nature of the diamond, which could be much higher than that of a plain track. When trains move over changing track stiffness and geometry, wheel-rail interaction induces high-magnitude impact forces. As a result, it is recommended that such forces be accounted for in the design standards to prevent further failure, as well as financial loss and possibly loss of life.

During such time, train speed limit restrictions were put in place around the cracked bearer to prevent further damage to adjacent components and to minimize any risk of derailment. The speed limit at this track location was limited to 50 km/h due to the turnout degradation. Apart from that, it was also found that the condition of the thick SA-47 pads is still in very good condition. No severe abrasion on either bearers or pads can be observed.

3.2 GEOTECHNICAL INSPECTION

Geotechnical Services carried out an inspection and investigation by test pits at the area during a track possession. [Figure 6.4e and f](#) shows the cross profile of a test pit at the country-end v-crossing. From the geotechnical investigation, the ballast,

**FIGURE 6.4**

Crack of the concrete bearer under the rail pad. (a) Geometry of Clyde 700 diamond (looking toward Sydney). Poor line smoothness (lateral alignment) and vertical rail dip can be observed at the diamond. (b) Ballast pulverization at k-crossing, black greasy material was collected at fastening area. Note that there is a small gap between ballast and the ends of bearers subject to high-impact force due to some small dilation of ballast. (c) Ballast pulverization at country-end v-crossing. (d) Crack of the concrete bearer under the rail pad. (e) Cross profile of ballast and subgrade at the country-end v-crossing at rail seat area. Ballast contamination can be observed. It is noted that there is a high level of ballast densification due to high-frequency loading. (f) Cross profile of ballast and subgrade at the country-end v-crossing at middle span of bearer (1 m deep excavation).

capping layer, and formation conditions within the area were considered reasonable. The ballast layer depth at this location is about 285 mm on top of the dense capping layer. Certain level of fine contamination in ballast and its densification could be observed and the subsoil drainage was deemed not to be very good. However, there was no sign of distress and failure that could be observed from the substrata.

4 CURRENT ANALYSIS AND DESIGN METHODS

Turnout bearers undergo a lot more stress than conventional sleepers in railway tracks, as the train is moving from one track to another. Standards have been set in Australia regarding the design load on concrete sleepers, which can be found in AS 1085 Part 14, and for RailCorp in SPC 233 [16–19]. The Australian Standards 1085, part 14 (AS 1085.14) sets out the requirements for design, manufacture, testing, and installation of prestressed concrete sleepers and fastenings for railways across Australia [16]. Under its Appendix A, the standards outline the details of turnout bearers, the manufacturing details, loadings and design, fastenings, and testings. Turnouts consist of bearers of various lengths, with rails attached at predetermined locations to allow tracks to be connected to adjacent tracks. Crossovers may have discontinuous bearers along the connection, depending on how far apart the adjacent tracks are to each other. In addition to the standards set out in AS 1085.14 for standard sleepers, further considerations must be made for the extra dynamics of turnouts, including:

- (i) Same method for standard bearers shall be used for *distribution of axle load*;
- (ii) The dynamic loading on the bearer shall be designed for at least 2.5 times the static load for the impact factor;
- (iii) Effects of centrifugal force should be allowed for curved rails;
- (iv) Forces and moments shall be considered from point motors as well as other equipments required;
- (v) Special load distributions specified by the designers;
- (vi) Generally, bearers extend further than standard sleepers, and this must be accounted for in the ballast layout. In cases where they extend beyond the ballast shoulders, the moments and shears should be calculated assuming the ballast and subgrade behave as elastic foundation;
- (vii) In the case of negative bending due to the extended bearer lengths, there must be adequate bending capacity built into the bearers to deal with the dynamic shear forces and bending moments. The bearer should be designed to deal with negative bending that is at least two-thirds of the maximum positive bending moments;
- (viii) Transverse reinforcement should be provided at the bearer ends should there be substantial bending moments at the ends.

Based upon the AS 1085.14, the standard specifies the exact conditions and dimensions for concrete bearers. It has specified shear force and moment envelopes set out for the design of the bearers, as outlined in [Figures 6.5–6.8](#) [17].

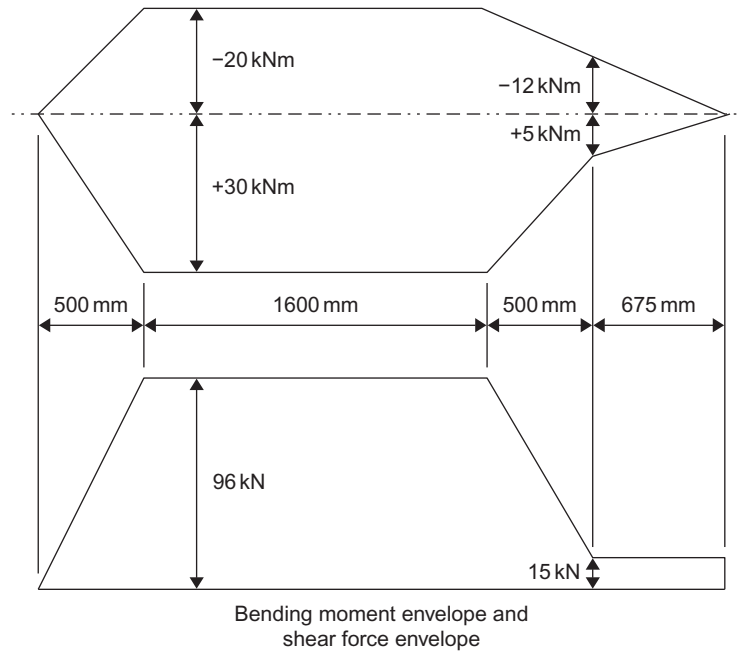


FIGURE 6.5

Shear force and bending moment envelopes for point motor bearer.

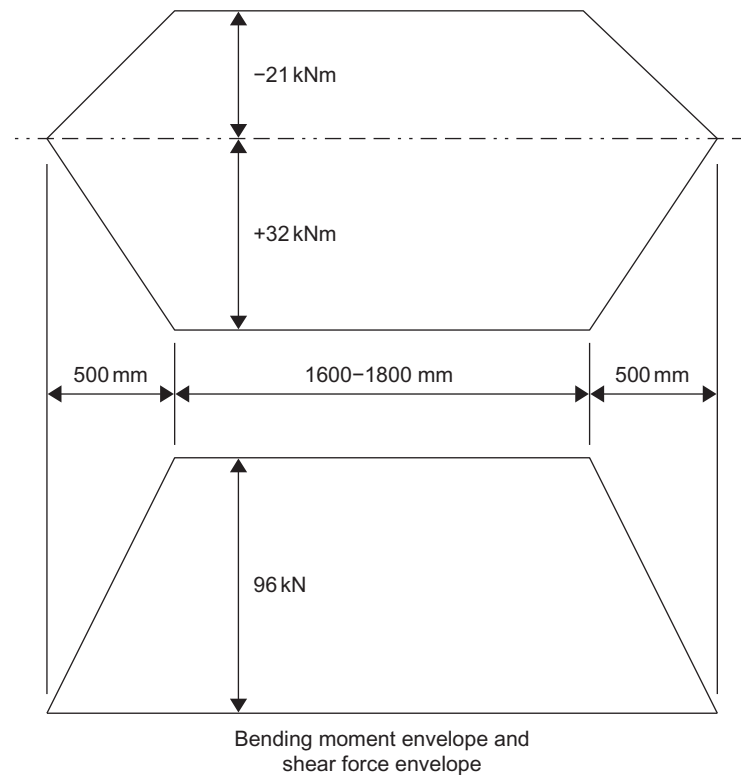
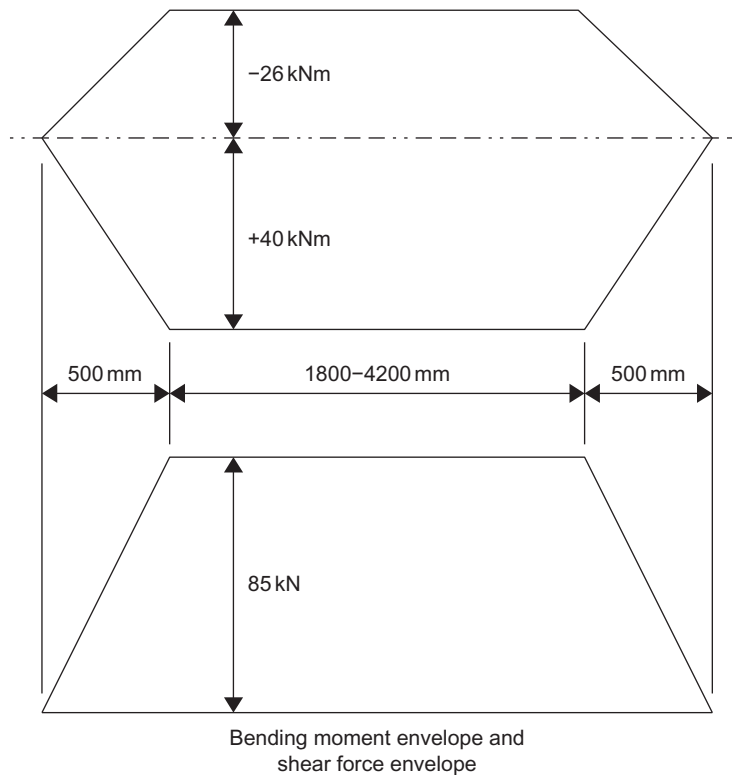


FIGURE 6.6

Shear force and bending moment envelopes for bearer lengths 2.6-2.8 m.

**FIGURE 6.7**

Shear force and bending moment envelopes for bearer lengths 2.8-5.2 m.

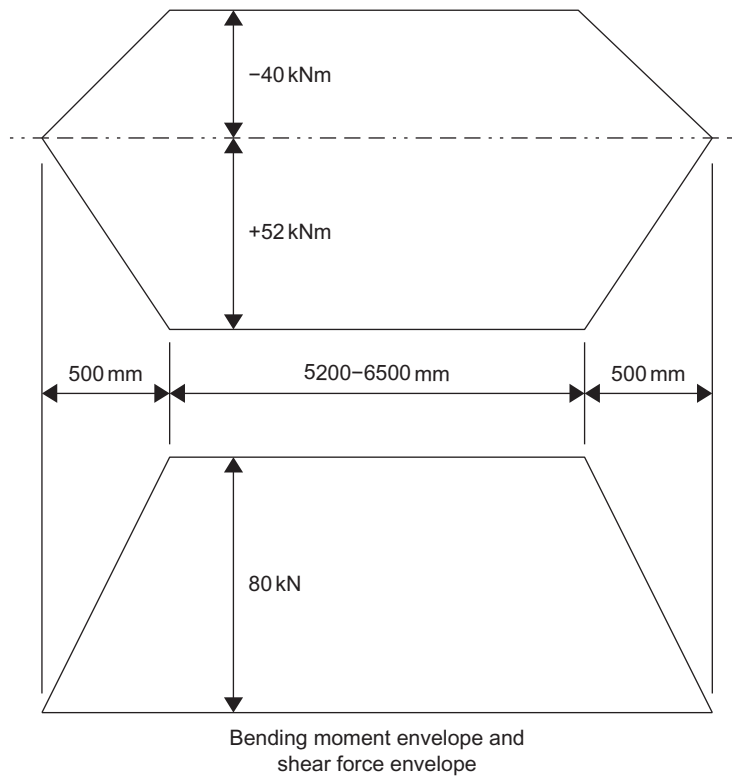
4.1 METHODS OF ANALYSIS

Actual contact pressure distribution in sleepers and bearers varies over time, making it virtually impossible to predict the exact conditions in actual conditions. Various hypothetical contact pressure distributions are shown in Figure 6.9. In Figure 6.10, the effect of the modulus (K_s) of ballast shows that the stiffer the sleeper support (the ballast), the more localized the contact stresses will be. Further, in Figure 6.11, the more flexible the sleeper is, again the more localized the contact stress will be.

4.1.1 Determination of rail seat load

The magnitude of the load applied to the sleepers from the rail depends on these known and unknown figures:

- (i) the rail weight;
- (ii) the sleeper spacing;
- (iii) the sleeper stiffness characteristics;
- (iv) the track modulus per rail;
- (v) pad stiffness;

**FIGURE 6.8**

Shear force and bending moment envelopes for bearer lengths 6.2–7.5 m, which span parallel tracks in crossovers.

- (vi) the amount of lateral movement between the rail and the sleeper;
- (vii) the amount of lateral movement between the sleeper and the ballast.

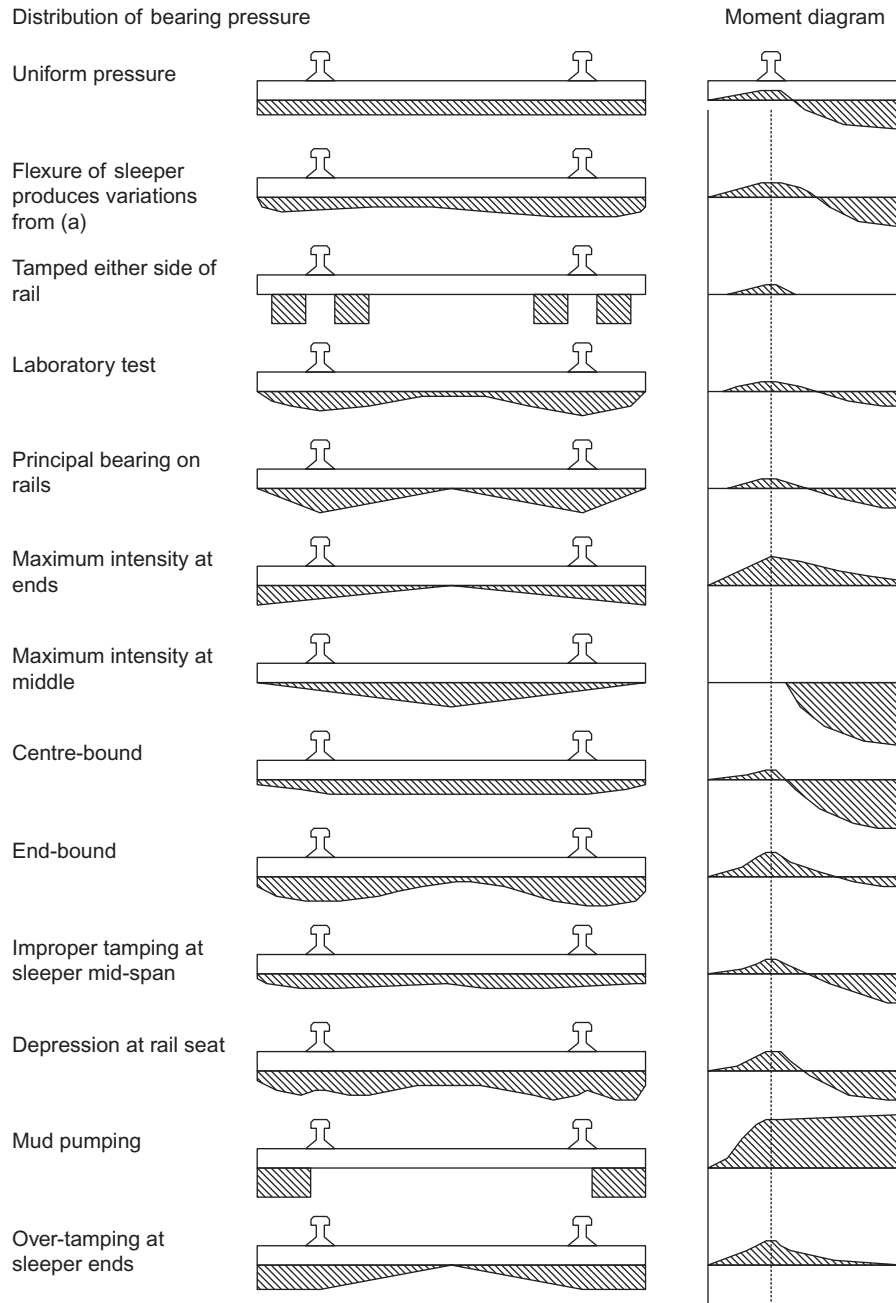
Various methods have been developed to analyze the rail seat load, and we will look at those used most frequently below.

4.1.2 The BOEF model

Under this model, suggested by Talbot and Clarke [20], one can determine the maximum rail seat load by:

$$q_r = S \cdot k \cdot y_m \cdot F_1,$$

where q_r is the predicted rail seat load; S is the sleeper spacing; k is the track modulus; y_m is the the maximum rail deflection caused by the interaction of a number of axle loads about a given reference position (Figure 6.12); and F_1 is the factor of safety to account for the variations in the track support caused by variations in the standard track maintenance (Clarke adopts F_1 is the 1).

**FIGURE 6.9**

Hypothetical distribution of bearing pressure and bending moment.

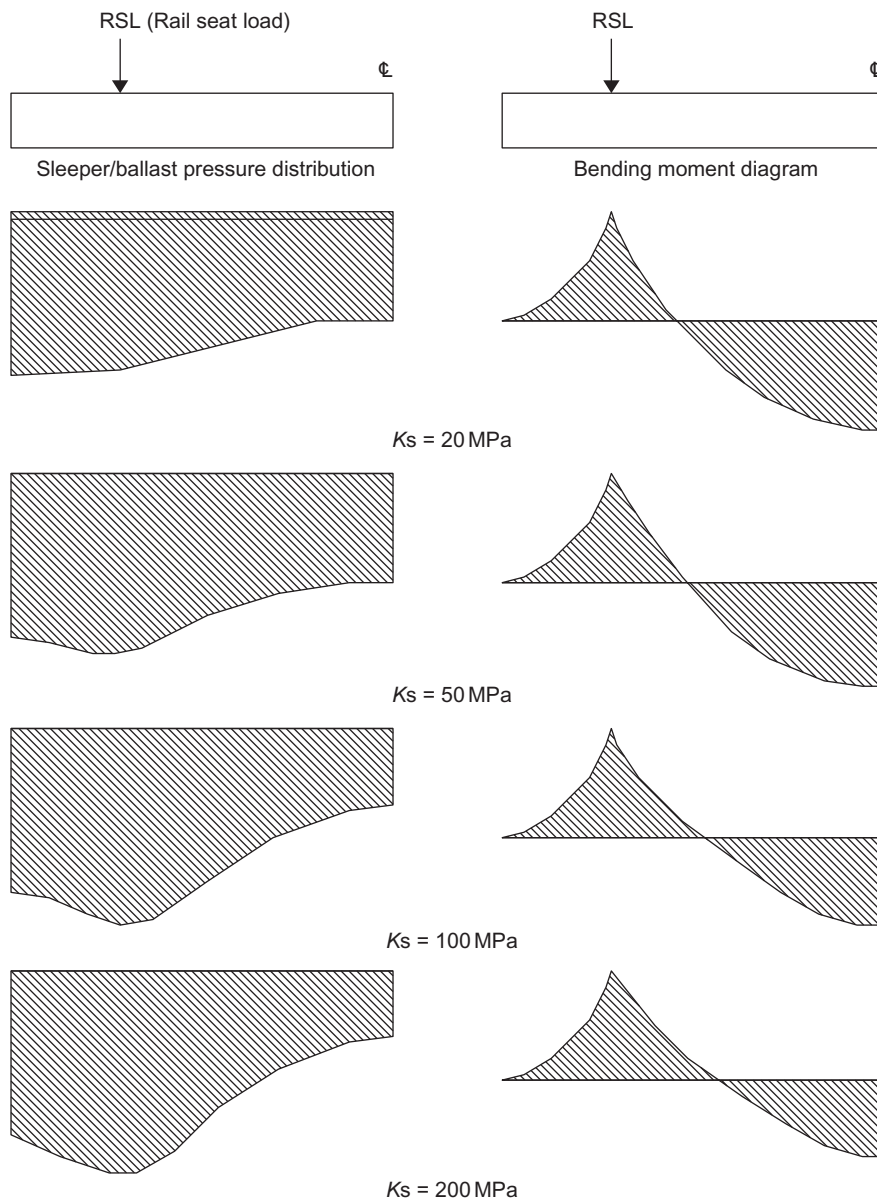
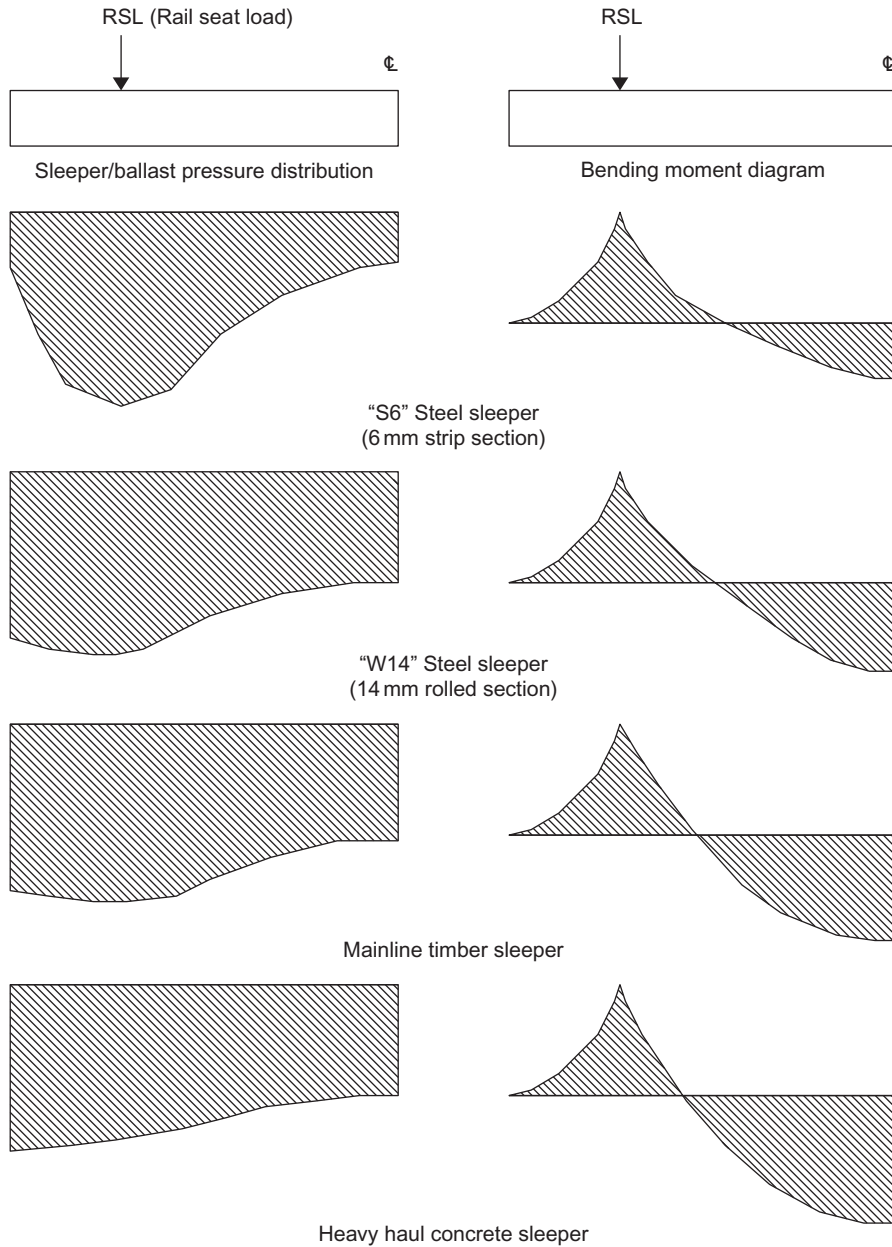


FIGURE 6.10

Sleeper support modulus on ballast pressure and sleeper bending moment.

**FIGURE 6.11**

Sleeper support stiffness on ballast pressure and sleeper bending moment.

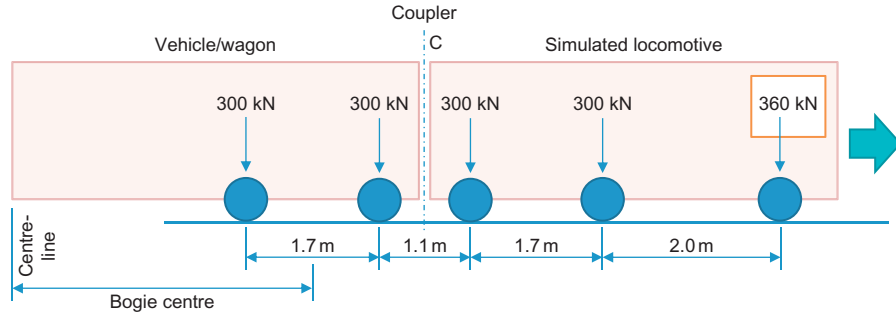


FIGURE 6.12

General load interaction diagram.

4.1.3 The American Railroad Engineering Association method

American Railroad Engineering Association (AREA) developed a formula based implicitly on the elastic foundation model [20]. All experimental work used a constant spacing of 510 mm between the sleepers, with an assumed track modulus of 13.8 MPa. Therefore, the equation for the track spring constant for all bearers was defined as:

$$D = k \cdot S \times \frac{A_s}{228 \times 10^3} = \frac{13.8 \times 510}{10^3} \times \frac{A_s}{228 \times 10^3} \approx 3 \times 10^{-5} A_s,$$

where D is the track spring constant; k is the track modulus; S = sleeper spacing; and A_s is the effective sleeper support area beneath the rail seat (mm^2).

Based upon further experiments, it can be determined that the product of the track modulus and maximum deflection, ky_m , to be nearly constant, and hence the effective area has little effect on the rail seat load. AREA then developed a simplified diagram (Figure 6.13), and showed that the AREA method could be expressed by:

$$q_r = D_f \times P,$$

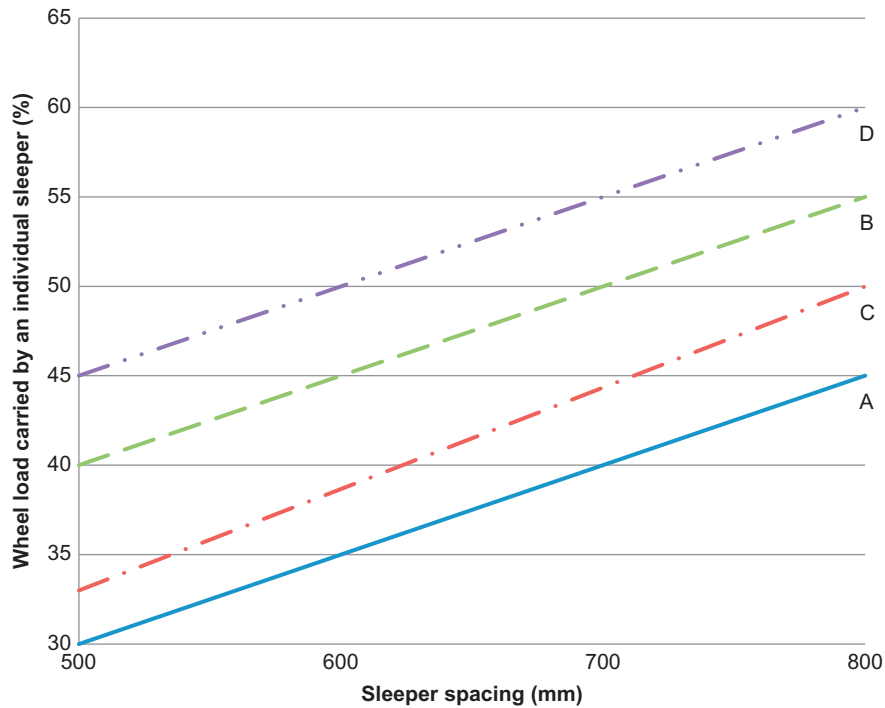
where q_r = maximum rail seat load (kN); P = design wheel load (kN); D_f = distribution factor, expressed as a proportion of the wheel load.

4.1.4 The ORE method

The Office of Research and Experiments of the International Union of Railways (ORE) developed a statistical method to calculate the maximum rail seat load [20]. Through experimental analysis, a formula was derived determining the maximum rail seat load:

$$q_r = \bar{\epsilon} \cdot c_1 \cdot P,$$

where P is the design wheel load, based upon the ORE formula for the impact factor; $\bar{\epsilon}$ is the dynamic mean value of the ratio q_r/P_s where q_r and P_s are mean values of the rail seat load and the static axle load, respectively, $c_1 = \frac{\bar{\epsilon}}{\epsilon}$, where ϵ is the the maximum value of the ratio q_r/P_s and c_1 is $\cong 1.35$.



Legend:

- A Estimate for timber sleepers (Magee, 1965-1971)
- B Area estimate for timber sleepers (AREA, 1975)
- C Adopted for steel sleepers (a) (Tew & Marich, 1988)
- D Area estimate for concrete sleepers (b) (AREA, 1985)

Note:

- (a) Based on a track modulus of 20 MPa and a rail size of 50 kg/m (note: a 25% allowance is applied to account for the effects of wheel interaction).
- (b) Area adopted a 5% increase above the area timber sleeper estimate due to the greater mass and stiffness of concrete sleepers.

FIGURE 6.13

Approximate percentage of wheel load carried by an individual sleeper [22].

Measured values of $\bar{\epsilon}$ can be found for various sleepers in Figure 6.14.

4.1.5 The three adjacent sleepers method

Based upon a simple assumption [20], the maximum possible sleeper loading occurs when the wheel load is distributed between three adjacent sleepers. The central sleeper will have a load of x percent, while the two adjacent ones will have

Sleeper type	Sleep dimensions length × breadth × rail seat thickness (mm)	Sleeper spacing (mm)	Area of sleeper support (10 ³ mm ²)	Q (mm)	$\varepsilon = \frac{\bar{q}_r}{\bar{P}_s}$
French, type VW Prestressed concrete	2300 × 250 × 140	600	200	400	0.56-0.59
British, type F Prestressed concrete	2515 × 264 × 200	760	268.8	510	0.32-0.44
German, type 58 Prestressed concrete	2400 × 280 × 190	600	252	450	0.32-0.44
French, Hardwood	2600 × 255 × 135	600	280.5	550	0.46-0.76

The area of sleeper support = $2Q \times$ sleeper width at rail seat, where Q = distance of the centre line of the rail from the end of the sleeper.

FIGURE 6.14

Values of $\bar{\varepsilon}$.

a value of $x/2$ percent. Therefore, the value of x will be 0.5, and hence the maximum rail seat load is:

$$q_r = 0.5P,$$

where P is the design wheel load; q_r is the predicted rail seat load.

4.2 SLEEPER BENDING MOMENTS

The material chosen for the analysis of maximum sleeper bending at the rail seat and center is irrelevant, but rather the flexural limits, usually expressed in the form of positive or negative bending moment capacity in concrete, imposed on the sleeper varies accordingly.

4.2.1 Maximum sleeper bending moment at the rail seat

For determining the maximum bending, a uniform effective contact pressure distribution between the sleeper and the ballast is assumed [20] and the calculations based upon this simple assumption would yield an upper-bound solution sufficient for sleeper design purposes.

In the upper-bound solution, it is assumed that the support from the ballast is a point load at the end of the sleeper, equal to the rail seat load (Figure 6.15a), giving the equation:

$$M_r = q_r \left(\frac{l-g}{2} \right),$$

where M_r is the maximum sleeper bending moment at rail seat; q_r is the maximum rail seat load; l is the overall sleeper length; and g is the distance between the rail centers.

A more realistic method assumes half the rail seat load is distributed over the sleeper overhand length (Figure 6.15b), giving:

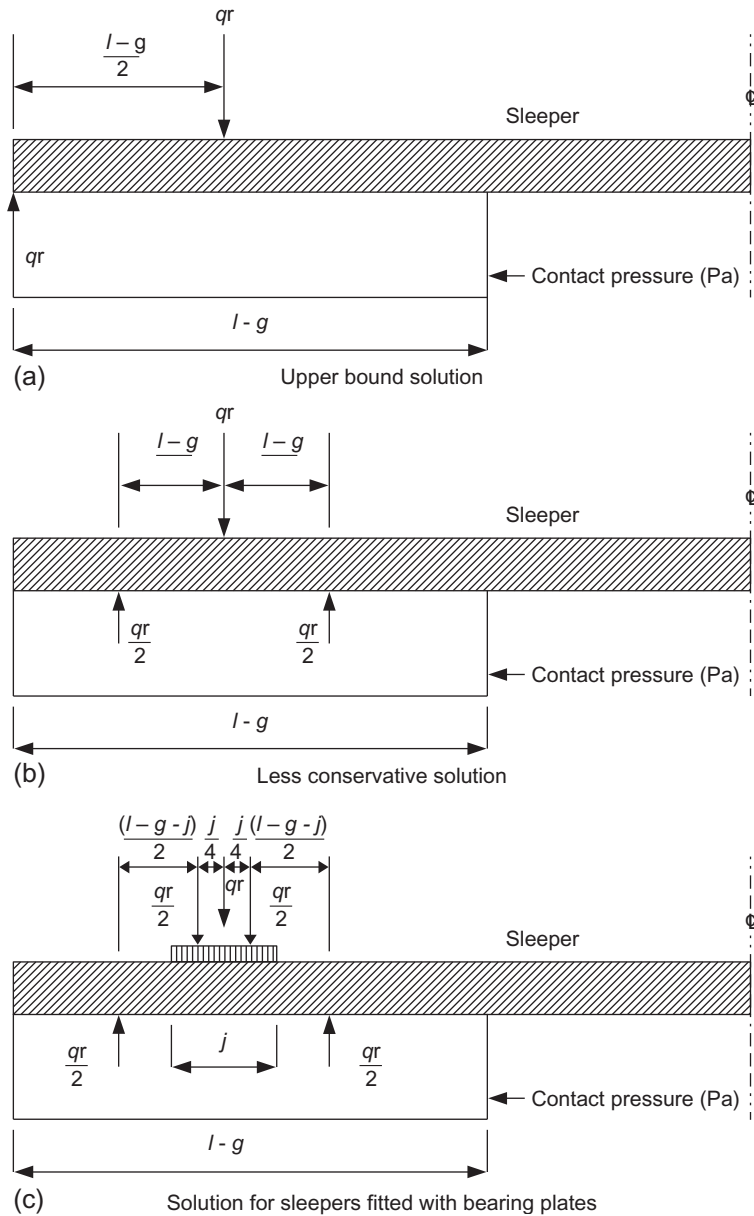


FIGURE 6.15

Pressure distributions for the calculation of design bending moment at rail seat [40–41].
(a) Upper-bound solution. (b) Less conservative solution. (c) Solution for sleepers fitted with bearing plates.

$$M_r = q_r \left(\frac{l-g}{8} \right).$$

For sleepers with bearing plates fitted (Figure 6.15c), then the maximum bending moment at the rail seat for this condition is given by:

$$M_r = q_r \left(\frac{l-g-j}{8} \right),$$

where j is the length of sleeper bearing plate.

4.2.2 Maximum sleeper bending moment at the center of the sleeper

Maximum bending at the center of the sleeper generally occurs when it is said to be centerbound (or when load is greater in the center). The two common solutions for this problem are the Battelle Solution and the Raymond Solution. Battelle assumes that the sleeper is resisted by a single-point load at the center of the sleeper, and can be calculated from:

$$M_c = q_r \frac{g}{2},$$

where M_c is the maximum sleeper bending at sleeper center.

This solution is believed to be extremely unlikely to occur in real-world situations, but can still be considered an upper-bound solution for the maximum sleeper bending moment at the center of the sleeper.

Raymond assumes a uniform bearing pressure over the total length of the sleeper, based upon experimental work carried out by Talbot [20]. Therefore, the maximum bending moment can be calculated from:

$$M_c = q_r \left[\frac{g}{2} - \frac{l}{4} \right] = q_r \left[\frac{2g-l}{4} \right].$$

4.2.3 Flexural requirements of sleepers

As noted earlier, the type of flexural limit imposed upon sleepers varies according to the sleeper type. For prestressed concrete sleepers, the design bending moment capacity of the sleeper is what must be considered. There are various methods of analysis, as outlined below.

4.2.3.1 The AREA design method

The AREA method assumes that, due to a gradual redistribution over time, the sleepers will settle on the ballast to a uniform contact distribution over the entire length of the sleeper [20]. This produces positive flexure at the rail seat and negative flexure at the center of the sleeper. Based on its own method of determining the rail seat load, the maximum positive bending moment at the rail seat is given by:

$$M_r = \frac{W(l-g)^2}{8},$$

where M_r is the maximum positive sleeper bending moment at rail seat; l is the total sleeper length; g is the distance between rail centers; W is the assumed uniformly distributed load.

The maximum negative bending moment at the center is given by:

$$M_c = \frac{Wg^2}{8} - M_r,$$

where M_c is the maximum negative bending moment at the center of the sleeper.

By combining these two equations, the solution is

$$M_c = q_r \frac{(2g - l)}{4}.$$

This moment calculated is considered adequate under normal service conditions, but sleeper cracking will still occur due to abnormally high loads.

4.2.3.2 The ORE design method

The ORE design method is based upon experimental data from both laboratory and actual track conditions. It also assumes the maximum rail seat load to be that calculated with the ORE method [20].

The ORE experiments found that the actual stresses are higher than the theoretical stresses, and were also location dependent. A parameter was chosen to adjust for the difference, being the *lever arm*. Also, the load spread is assumed to be 45° through the sleeper from the rail seat (Figure 6.16). Therefore, the lever arm is:

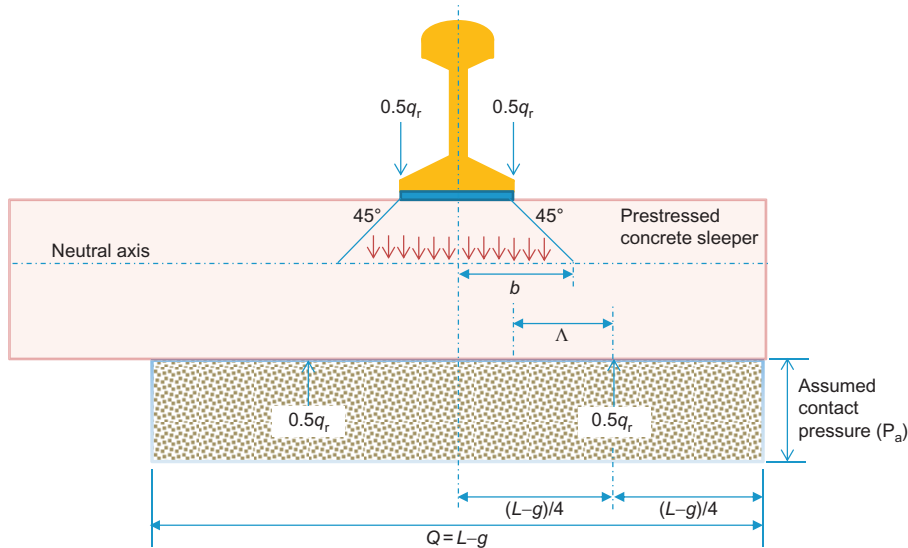


FIGURE 6.16

Determination of lever arm.

$$\Lambda = \frac{Q - b}{2},$$

where Λ is the lever arm; Q is the sleeper end distance; and b is the half length of rail loading at sleeper neutral axis.

The theoretical bending moment is calculated with:

$$M_r = \frac{q_r}{2} \Lambda,$$

where M_r is the rail seat bending moment.

To account for the theoretical and experimental moments measured, the lever arm is factorized:

$$\Lambda_m = \phi \cdot \Lambda,$$

where Λ_m is the factorized lever arm and ϕ = dynamic impact factor of the sleeper bending moment at the rail seat.

The empirical coefficient ϕ is found to be ~ 1.6 , and combining all these equations provides the maximum sleeper bending moment at the rail seat:

$$M_r = \bar{\epsilon} \cdot c_1 \cdot \phi \cdot \frac{\Lambda}{2} \cdot P.$$

Similarly, the bending moment at the center can be determined to be:

$$M_c = M_r \cdot \zeta \cdot \frac{EI(\text{centre})}{EI(\text{railseat})},$$

where M_c is the bending moment at sleeper center; ζ is the scale factor relating bending stress at rail seat to sleeper center; and EI is the sleeper stiffness.

4.2.3.3 Australian design code

The Australian design code is in many ways similar to the AREA method, but is more rigorous [20]. The positive moment, with load distributed as shown in Figure 6.17a, at the rail seat is given by:

$$M_{r+} = \frac{q_r(l - g)}{8},$$

where M_{r+} is the maximum positive bending moment at rail seat; q_r is the design seat load; l is the sleeper length; g is the distance between rail centers.

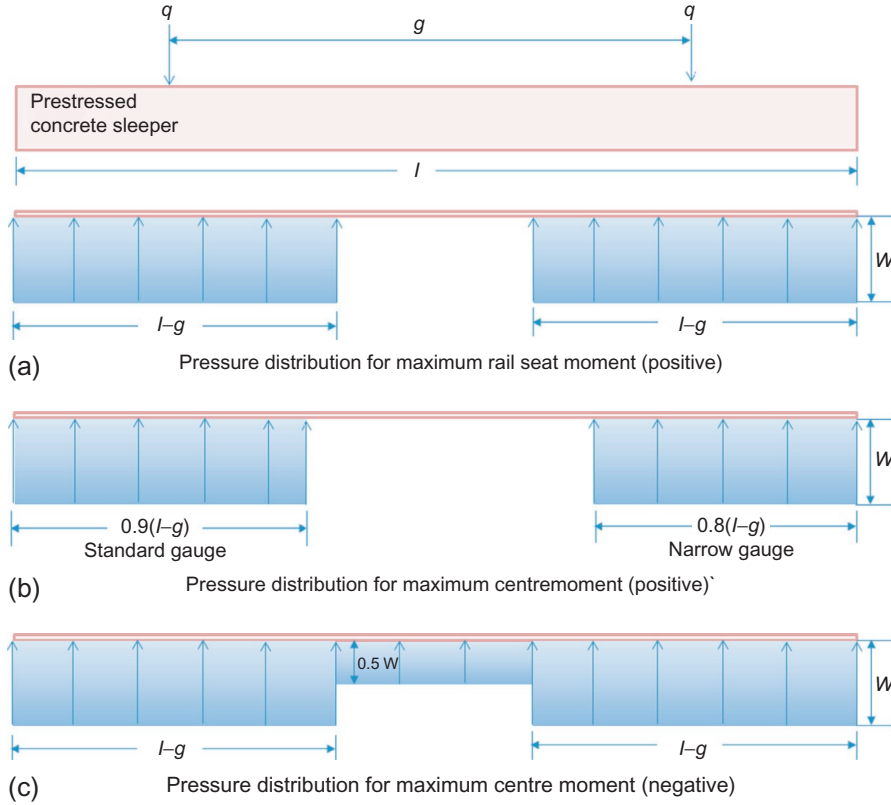
The maximum negative bending moment is given by either 67% of the maximum positive moment, or 14 kN m, whichever is greater.

For the center of the sleeper, the maximum positive moment, with load distributed as shown in Figure 6.17b, is given by:

$$M_{c+} = 0.05q_r(l - g), \text{ for standard gauge,}$$

$$M_{c+} = 0.10q_r(l - g), \text{ for narrow gauge,}$$

where M_{c+} is the maximum positive bending moment at the sleeper center.

**FIGURE 6.17**

Assumed contact pressure distributions for concrete sleeper design. (a) Pressure distribution for maximum rail seat moment (positive). (b) Pressure distribution for maximum center moment (positive). (c) Pressure distribution for maximum center moment (negative).

The maximum negative bending moment, with load distributed as shown in Figure 6.17c, for the center is given by:

$$M_{c-} = 0.5 \cdot \left[q_r \cdot g - W \cdot g(l-g) - W \frac{(2g-l)^2}{8} \right],$$

where M_{c-} is the maximum negative bending moment at sleeper center; $W = \frac{4q_r}{3l-2g}$ (maximum load per unit length of sleeper).

4.3 CURRENT DESIGN METHOD

4.3.1 Essential loading determination

To begin the analysis, basic loading diagrams must be analyzed. Determining the load experienced by the rail seat requires the usage of the methods mentioned earlier, being the BOEF model, AREA method, ORE method, and the three adjacent sleepers method, and comparing the results [20]. Using the discussed analysis methods, the load

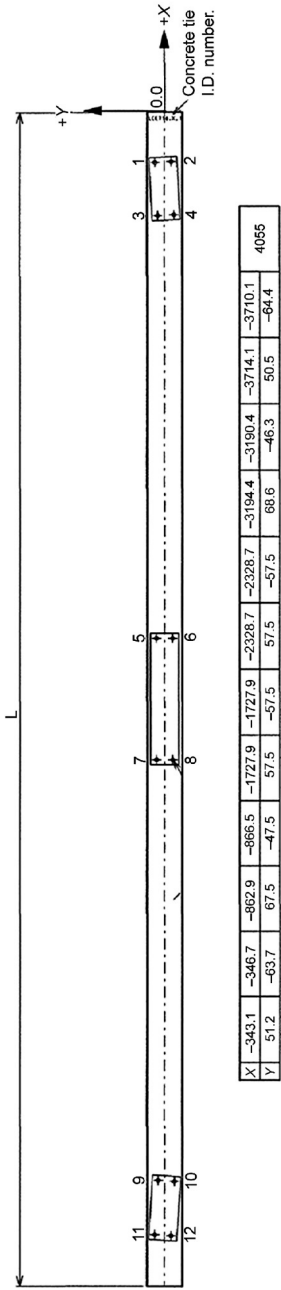
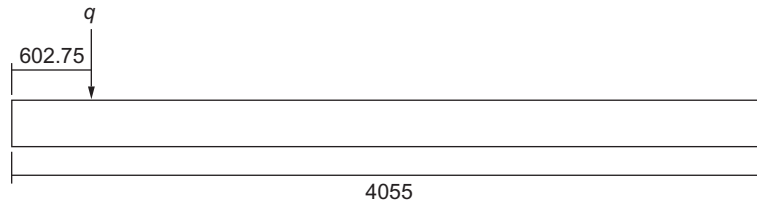


FIGURE 6.19

Concrete tie layout detail.

**FIGURE 6.20**

Loading diagram.

This leads to the basic loading diagram of the bearer in question, shown in Figure 6.20. There is a track stiffness which provides support underneath the bearer along its length. The distance from the bearer end is an approximation, as the rail seat also accommodates the check rail, which is designed to keep the wheel on the track during the transition (note that all figures are in millimeters).

5 FINITE ELEMENT MODELING OF TURNOUT DIAMOND

5.1 COMPLETE DIAMOND ANALYSIS

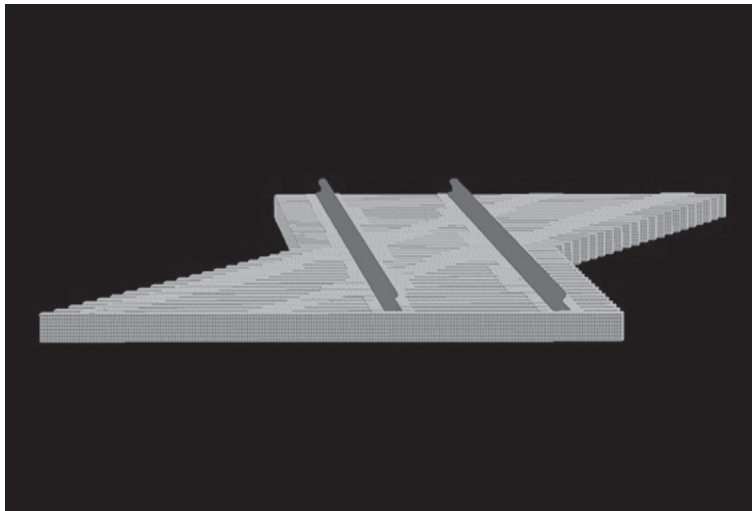
A finite-element model of turnout diamond has been developed using a finite-element package, STRAND7. Analysis of the complete diamond requires modeling each bearer in the whole crossover, on which rail is applied, as shown in Figure 6.21.

Determining the load points on the rail requires a review of rolling stock data to discover the train with the closest loading points (assuming the closer the axles are, the more the forces will interact). Figure 6.22 shows the typical train bogey and axle setup. Assuming the axle load is equal throughout, the bogeys are closest at the carriage joints, as opposed to the singular carriage itself (i.e., $2 \times \text{body overhang} < \text{body centers}$). The actual data from each rolling stock, shown in Table 6.3, shows that Train 1 has the shortest distance between the two furthest axles compared to the three other rolling stock carriages.

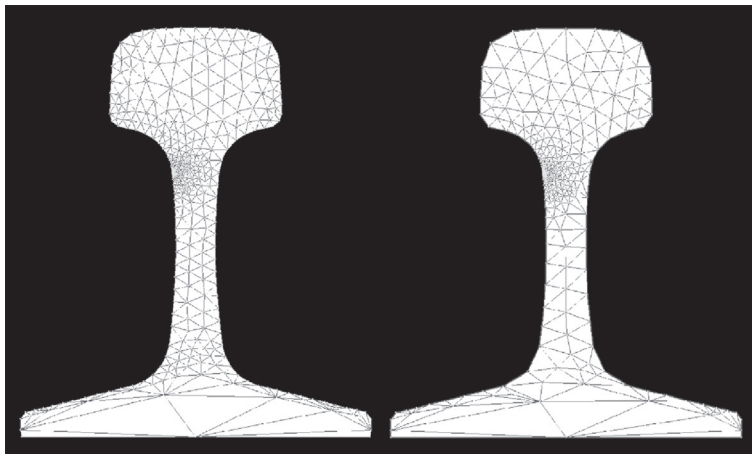
Figure 6.23 shows the half model with the point loads applied. The track stiffness on all the bearers is calibrated using the field data, which is the same as on the singular bearer analysis, 0.1689 MPa/mm. Again, it is assumed that the axle load above the failed bearer is supported by the single rail (noticeably with the larger load shown in the figure).

5.2 INDIVIDUAL BEARER ANALYSIS

The failure analysis of the individual bearer involves modeling the single bearer, and the rail pad and steel pad on which the rail sits, is shown in Figures 6.24 and 6.25 [43]. Again, the specifications can be obtained from the concrete tie layout detail, with the bearer cross-section of 250 mm $\times$ 250 mm, as per RailCorp standard [16,17].



(a)



(b)

FIGURE 6.21

Complete crossover modeled with detailed model of rail on top. (a) Turnout diamond model. (b) More detailed rail model (left) and simpler rail model (right).

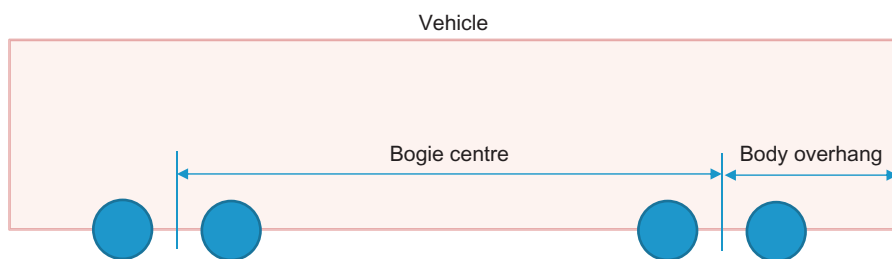
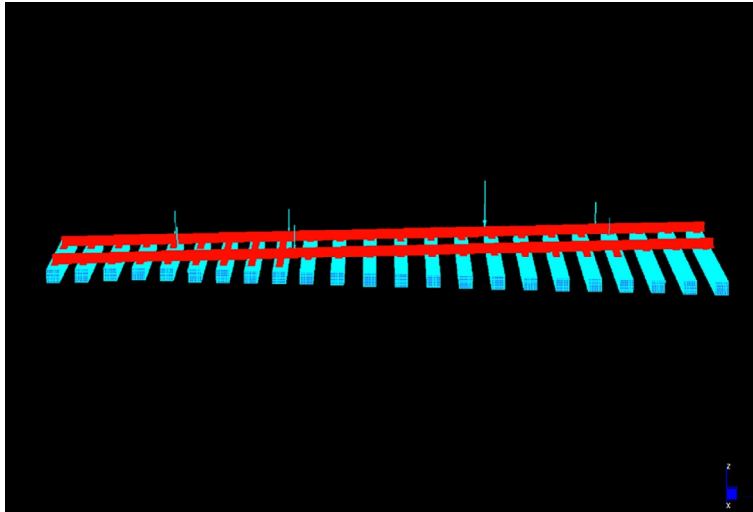


FIGURE 6.22

Typical carriage layout.

Table 6.3 Distance Between Bogey and Axles

Train	1	2	3	4
Axle distance	2300	2400	2440	2440
Bogey distance	3998	3876	4670	3831
Axle distance	2300	2400	2440	2440

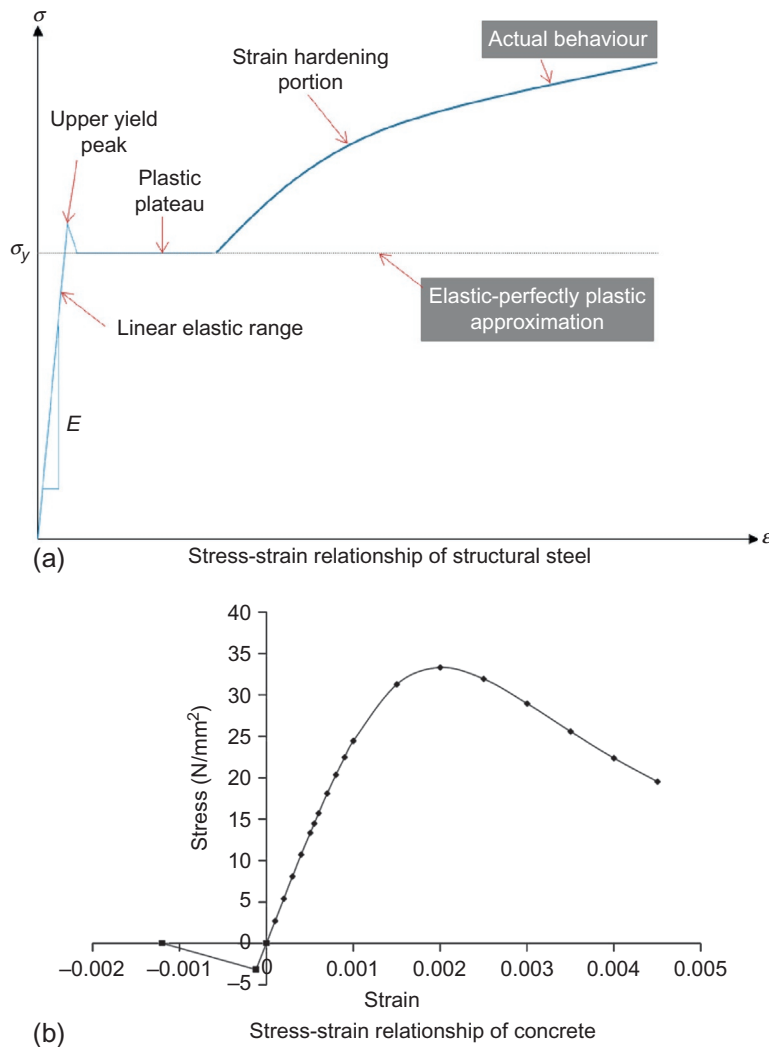
**FIGURE 6.23**

Half bearer with loads applied.

Applying the loads calculated earlier to the rail seat requires a conversion to a pressure over the area of the pad. To determine the contact area of the pad, the rail dimensions are used, with the foot width (146 mm) and rail pad width (115 mm), giving an area of 16,790 mm². This can give a pressure load for each of the methods mentioned earlier, as shown in [Table 6.4](#).

These have to be adjusted to accommodate the model, as the pressure area is slightly smaller, producing a slightly larger pressure, as shown in [Table 6.5](#).

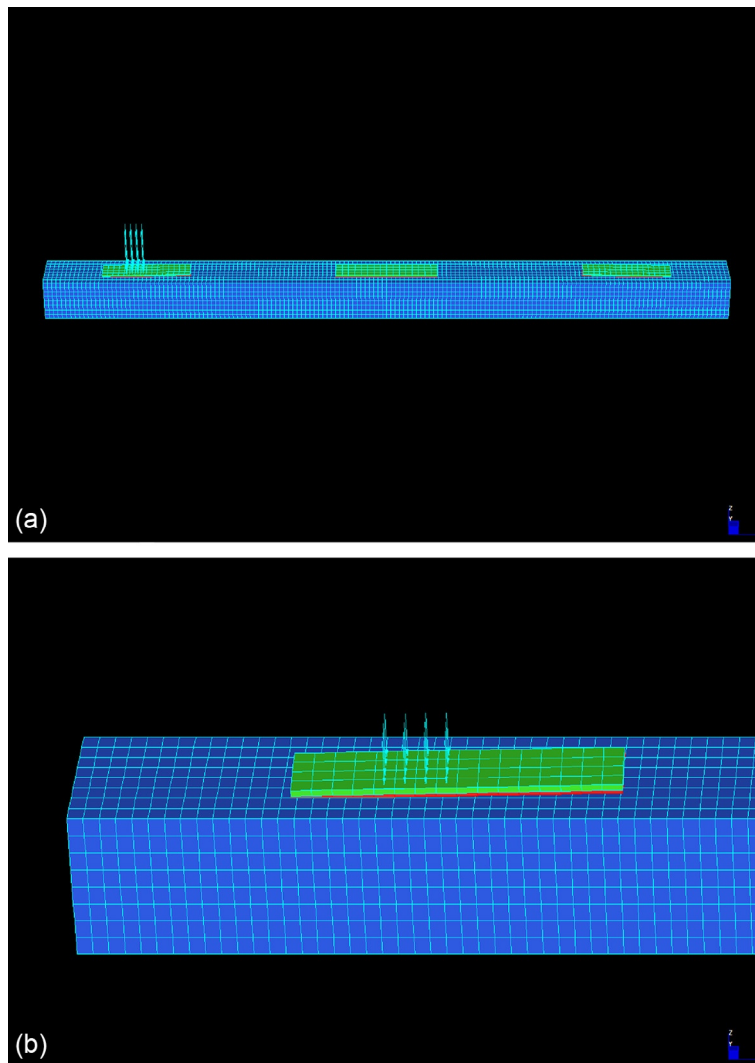
Track stiffness can be applied to the base of the model to provide support for the load and it has been derived from the specifications discussed earlier. This value has been converted to MPa/mm, and as it is uniform across the whole diamond, the minimum support required (i.e., the maximum value) is used in the calculations, being 0.1689 MPa/mm. The model validation has been carried out in comparison with field measurements (displacements and track modulus) to ensure that the model can accurately provide practical results.

**FIGURE 6.24**

Material models in STRAND7. (a) Stress-strain relationship of structural steel. (b) Stress-strain relationship of concrete.

5.3 CONTACT SURFACES

Using the contact surface element provided by STRAND7, as for the joining of non-aligned bricks, links have to be formed. This means that the load distributed through the nodes are not transferred to the other nodes, but rather spreads out on the whole

**FIGURE 6.25**

Finite-element modeling of turnout bearer. (a) Bearer model. (b) Loading on rail pads.

Table 6.4 Theoretical Loading Pressure for Respective Rail Seat Loads

Method	Rail Seat Load (kN)	Rail Seat Pressure (MPa)
Three adjacent method	147.5	8.785
BOEF model	103.25	6.149
AREA method	144.55	8.609
ORE method	156.35	9.312

Table 6.5 Actual Model Loading Pressure for Respective Rail Seat Loads

Method	Model Rail Seat Area (mm ²)	Rail Seat Pressure (MPa)
Three adjacent method	14,820	9.953
BOEF model	14,820	6.967
AREA method	14,820	9.754
ORE method	14,820	10.550

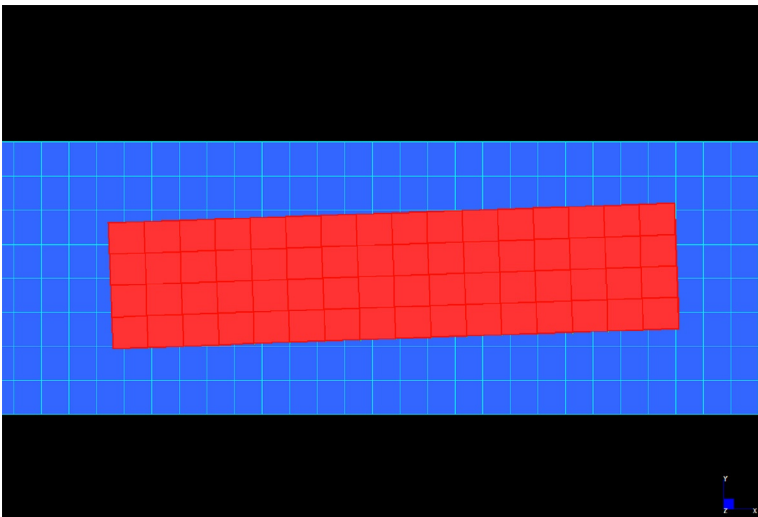


FIGURE 6.26
Attaching nonaligned nodes to brick faces, such as pads onto bearer.

brick face, which may have reduced the overall pressure. This contact element (as shown in [Figure 6.26](#)) enables two solid elements (bearers and pads) to contact and redistribute forces.

6 RESULTS AND DISCUSSIONS

6.1 INDIVIDUAL BEARER ANALYSIS

After running the model analyses, the model results showing how the load is spread along the rail pads onto the bearer, and also how the bearer is supported by the ballast underneath. The forces experienced by the models are shown in [Figure 6.27](#), below, with the maximum forces (FZ) shown at the top of the table.

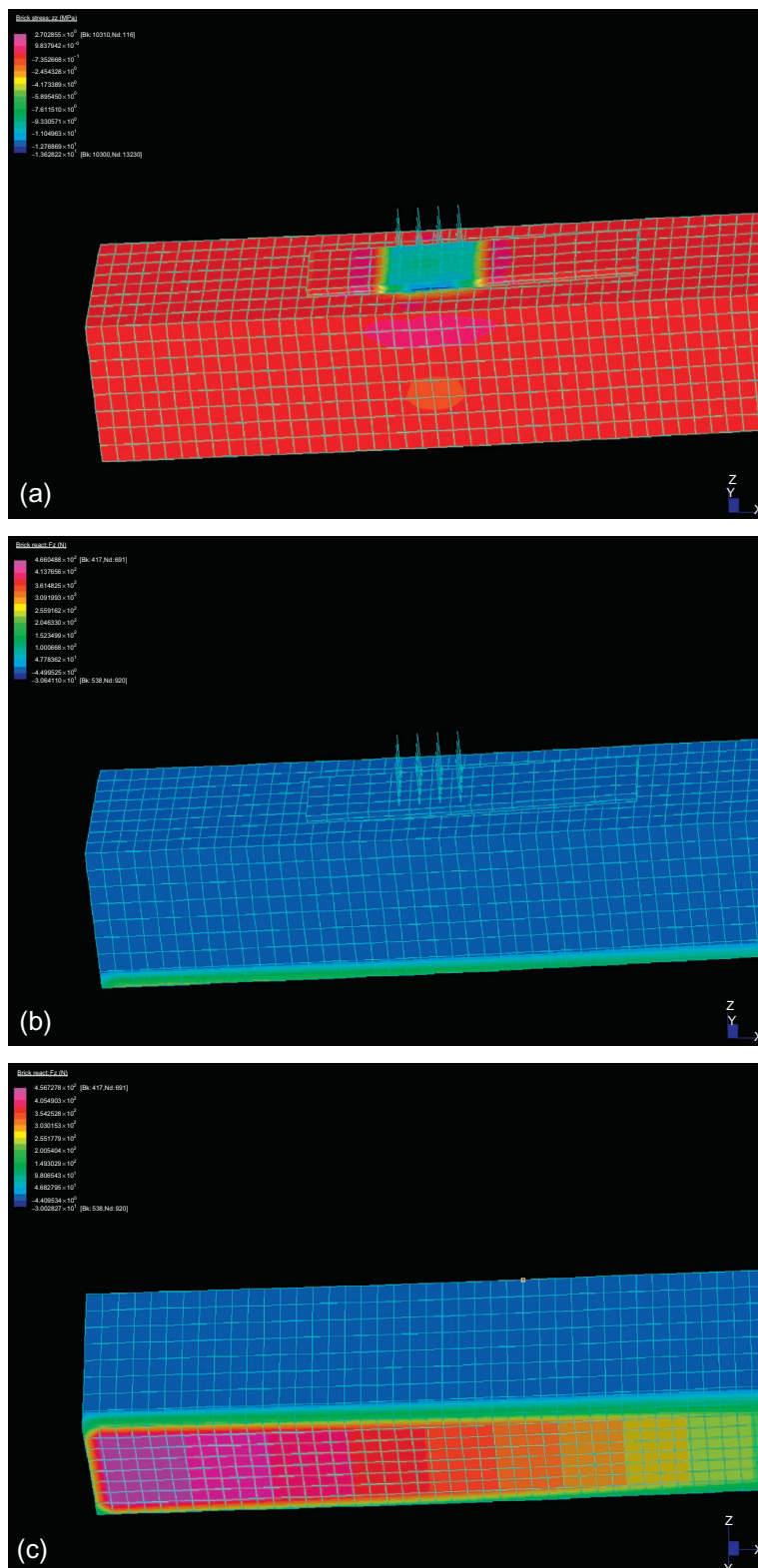


FIGURE 6.27

Finite-element analyze results of individual bearer. (a) Pressure distribution from loading. (b) Forces experienced from loading. (c) Forces experienced underneath bearer from loading.

6.2 COMPLETE DIAMOND ANALYSIS

The resultant model of the diamond crossover analysis can be found in [Figure 6.28](#). They are the loading results from the analysis. As can be seen from the numbers, the results correlate well with the individual bearer forces experienced, sitting in between the minimum and maximum resultant force figures, at 378N.

6.3 DISCUSSIONS

The reaction forces underneath the bearers show that as the load is increased, the bearer will definitely experience a greater force. [Figure 6.29](#) presents the sleeper/ballast contact pressures obtained from the methods of determining the rail seat load. It is found that the data is in very good agreement with those obtained from the more complete crossover model. It is found that the maximum resultant forces from the current methods of analysis are still far below the force envelope specified in the design standards. Even if it is to be increased substantially, the load would still not cause a failure of the bearer, according to the finite-element analyses.

This puts into question whether the diamond design contributed to the failure. One factor that was not taken into account for the analysis is the possibility of high-amplitude dynamic forces experienced by the rail and bearer. The dynamic impact forces would in turn apply a greater moment force upon the bearer. It is clear from [Figure 6.29](#) that if the rail seat loads are to increase only about 20-30%, the contact pressure between ballast and sleeper will have reached the allowable limit of 750 kPa [16,17]. Considering AS1085.14, if the specified impact factor of 2.5 is adopted, the resultant dynamic wheel load will gain 27% over the maximum allowable load (295 kN). It implies that potentially the ballast would significantly and rapidly settle under the crossings and diamonds. The large settlement can cause additional bending moment actions and their consequences, resulting in a flexural failure mode of bearer as observed in the field.

[Table 6.6](#) illustrates an estimation of turnout renewal costs when failure of diamond component incurs. It is important to note that criticality of turnout failure is relatively high as a turnout (only about 100-150 m in track length) tends to connect two or more plain tracks together. As such, its consequences are progressive and highly interconnected to the service values other connected tracks provide. This provides the fundamental rationale for higher priority of turnout renewal and maintenance than that of a plain track. From the cost estimate list, it can be seen that material costs cover around 6-7% of the total cost. It implies that re-designing critical structural components such as concrete bearers for the capability to resist higher impact loads could potentially save the associated penalty costs of 77%.

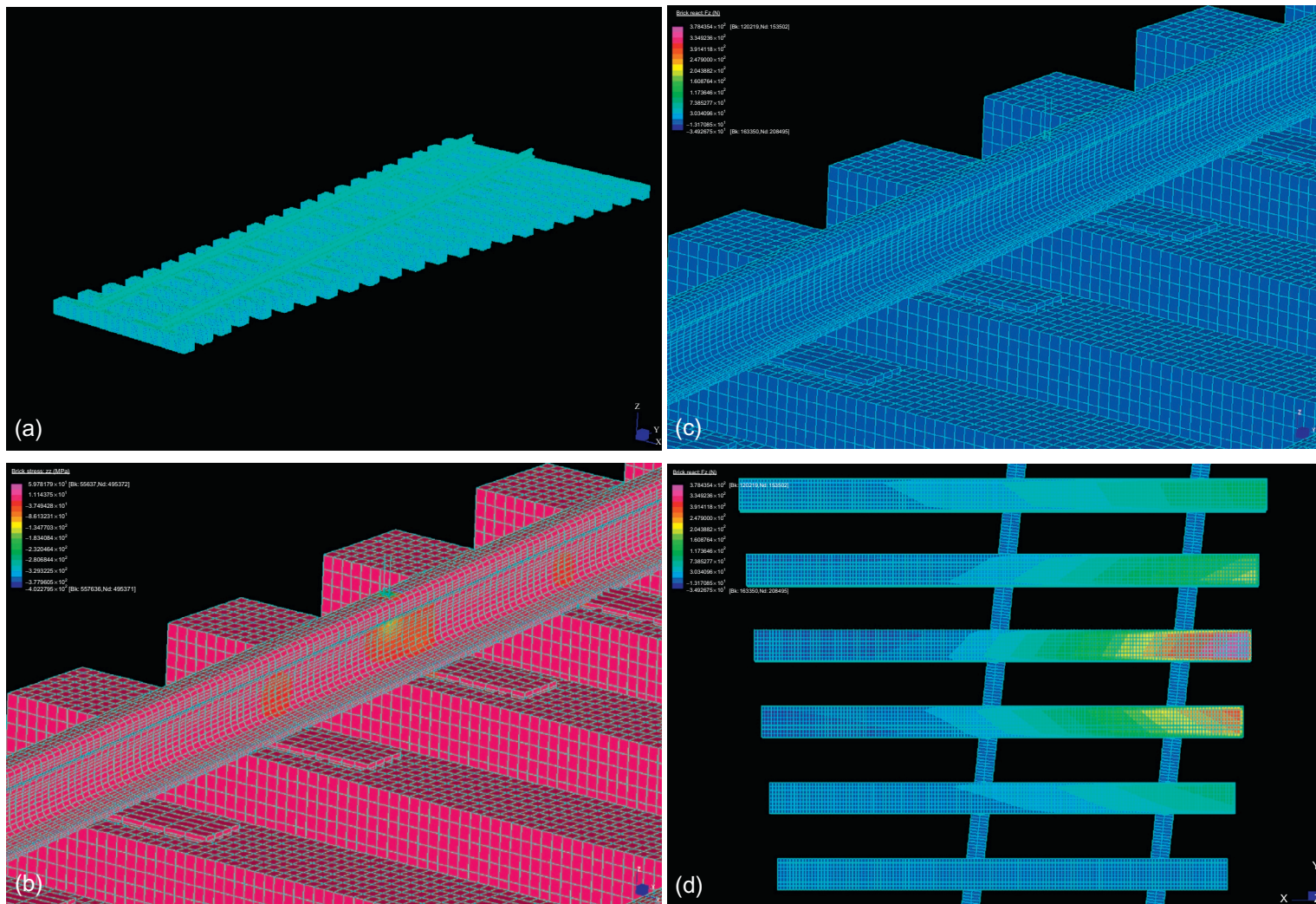
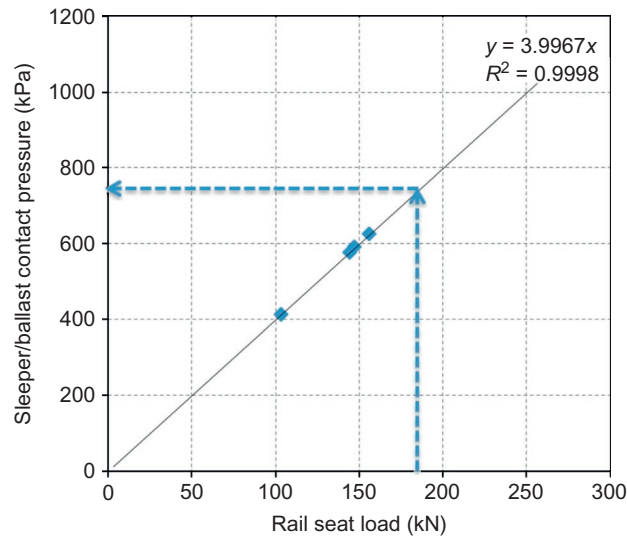


FIGURE 6.28

Finite-element analysis results of turnout diamond. (a) Crossover model. (b) Pressure distribution from loading. (c) Forces experienced from loading. (d) Forces experienced underneath bearers from loading.

**FIGURE 6.29**

Relation of forces and reactions using contemporary methods of analysis.

Table 6.6 Estimate Costs Associated with Turnout Renewal Due to Systemic Failure

Items	Renewal Cost (\$)	% Cost component
Materials	200,000	6.2
Machinery	75,000	2.3
Access/possession fee	50,000	1.6
Human resources and labors	150,000	4.7
Commission and signaling	100,000	3.1
Re-design, manufacturing	50,000	1.5
Formation improvement	100,000	3.1
Others (train replacement busses, logistics, delay costs, penalty fees, opportunity costs, etc.)	>2,500,000	77.5
Total	>3,225,000	100

7 CONCLUSION

Concrete sleepers/bearers are the main components supporting steel works and redistributing dynamic forces onto ballast and formation. Because of their low cost, high durability, ease in construction, and maintenance, concrete bearers will continue to be the popular components in modern turnout systems. However, wheel-rail contact over the crossing transfer zone has a dip-like shape and can often cause detrimental

impact loads on the railway turnout and its components. The field maintenance reports have demonstrated that a number of concrete bearers failed under repeated impact loadings, especially under turnout diamonds. Such an incident has caused major downtime, high cost of turnout renewal, and reduced reliability of train operations. As a result, a nonlinear 3D finite-element model has been developed, using commercial finite-element package STRAND7, to evaluate the design methods and to analyze the failure mechanism of the concrete sleepers/bearers at an urban turnout diamond. The finite-element modeling of the bearers and the diamond crossover has shown that the contemporary design methods may underestimate the vertical quasi-static loads for design of concrete bearers. The finite-element analysis results illustrate that the dynamic load action could be at or higher than the allowable sleeper/ballast contact pressure due to the variation of dynamic loading condition by 20–30%. It is highly likely that the variation could damage the supporting ballast and result in a premature flexural failure mode of concrete bearers. As a result, the outcome of this study recommends that a higher impact factor (>2.5) for the design of turnout bearers be adopted in railway industry. This adoption can potentially save millions of dollar from prioritized maintenance and unplanned renewal and from the penalty costs associated with the turnout diamond failure.

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Investigation of failure behavior of thin-walled tubular components and development of a procedure for evaluation of their mechanical and fracture properties

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1 INTRODUCTION

Thin-walled tubular components are widely used in various industries, for example, in the high-temperature and high-pressure regions of thermal power plants as boiler tubes, super-heater and re-heater tube banks, tubes of heat exchangers, steam generators, etc.

In nuclear reactors, thin-walled tubes are used as fuel-clad tubes as well as pressure tubes and calandria tubes, respectively. Zirconium alloys are widely used as the fuel-cladding tubes and other core internals in light-water reactors such as boiling water reactors, pressurized water reactors as well as pressurized heavy-water reactors.

Understanding of failure behavior of these fuel-clad tubes is important from safety point of view. Failure of fuel-clad can release enormous amount of radioactivity into the coolant and this in turn will contaminate other components in the reactor. In case leakage of coolant at flanges or other joints, the activity may be released to reactor containment. The decontamination of reactor containment and other systems in case of large-scale failure of fuel-clad tubes can add a huge economic penalty on the utility. Hence, assessment of clad-integrity is an important concern to operators of nuclear power plants as well as designers and safety analysts.

These zirconium-based alloys have favorable properties such as high mechanical strength, good ductility, very-low neutron absorption cross-section, and good corrosion resistance [1–3]. Fuel-clad tubes are of small thickness in order to have less resistance in the path of heat flow from the fuel to the coolant.

Under reactor operating conditions, the deformation behavior of fuel cladding is strongly influenced by the mechanical properties of the material. The mechanical properties also significantly change with years of reactor operation because of irradiation, high temperature, high stress, and corrosive environment, etc. Nonstandard tests (e.g., ring-tension test) on very small specimens need to be performed as it is difficult to extract standard specimens from these tubes. A ring tensile test (RTT) is one of the tests that use a ring specimen to obtain the hoop tensile properties of the tube-shaped structures.

The ring specimen has a curvilinear shape, and bending moment and deformation appear during the ring test. To reduce the bending moment, Arsene and Bai [4] proposed a central piece-inserted ring tensile specimen. Josefsson and Grigoriev [5] have used the ring-tension specimen to determine the mechanical properties of irradiated fuel pins. Limited bending stresses appear during pellet-clad mechanical interaction in the fuel-clad. The bending phenomenon rather affects the ring-tension test due to the specimen geometry, diameter difference between the ring and the mandrel, space (i.e., gap) between the semi-cylindrical mandrels, etc. [6,7]. The gauge geometry is not very well defined when plain rings are used in the tests. For this reason, many laboratories have developed RTTs with machined gauge section for the ring [4,6]. This specimen along with its modified versions and several other designs have been used in literature [8–19] to evaluate the ductility, creep and strain rate sensitivity, delayed hydrogen cracking velocity, and fracture resistance behavior of various types of Zircaloy fuel claddings.

It is known that the standard ASTM test techniques cannot be used for these thin and small-diameter tubular specimens. The results from these tests can also not be directly used to determine the mechanical properties and hence, a finite-element (FE) analysis is essential for this purpose. In this work, ring-tension tests were carried out on unirradiated Zircaloy-4 cladding tube specimens. Specimens for these tests were machined from the fuel pins in the form of rings. A split-mandrel was used to load these ring specimens in tension. Both the specimen and the mandrel were modeled to

evaluate the load-displacement behavior of the test. By comparing the load-displacement response of FE analysis with that of experiment, the material stress-strain curve was evaluated.

Investigation of fracture behavior of these fuel-clad tubes is of utmost importance to the designers and plant operators. It is important to ensure the structural integrity of these fuel-clad tubes under various types of operational transients as well as accident scenarios [4–9]. One of the postulated design-basis accidents for the fuel-clad tubes is the reactivity-initiated accident [9]. In such a situation, the control rod drops or ejects very rapidly from the reactor. This in turn deposits a large amount of energy in the fuel-pellets and leads to adiabatic heating and large fission gas release in the fuel pins. The fuel-pellets expand thermally and may cause fast straining of the surrounding Zircaloy clad tube through pellet-clad mechanical interaction.

The fracture behavior of these clad tubes is further complicated due to several conditions, such as hydrogen pickup, high temperature, irradiation environment, prior cold work, and heat treatment conditions, etc. In order to ensure that the clad remains intact for a longer period in the reactor, fracture-mechanics principles can be applied for the fitness-for-purpose of service assessment of these fuel pins. However, it may not be possible to evaluate the fracture behavior of these thin-walled tubular components using standard ASTM test techniques (ASTM Standard E 1820-13) [20]. This is because of the stringent requirements of the ASTM test methods [20] for satisfaction of plane strain condition at the crack-tip. For this purpose, several types of nonstandard specimens have been studied in literature [14,17–27] in order to evaluate the fracture resistance and transverse mechanical properties of thin-walled tubular components.

As standard fracture-mechanics specimens cannot be machined from these thin-walled tubes, nonstandard specimens with axial cracks (which can be directly machined from the fuel-clad tubes) have been used in this work for the fracture tests. Two types of loading mandrels have been used, viz., (a) a conical mandrel, driven through the axially cracked tubes simulating expansion of fuel-pellets inside the tubes and (b) two semi-cylindrical mandrels, inserted in the axially cracked tube and aligned in such a way as to open the diametrically opposite cracks. The effect of various values of initial crack length on the fracture resistance behavior of the tubes have been studied using these two different types of test setups simulating different types of loading conditions at the crack-tip of the specimens.

2 MATERIAL

In this work, the fracture behavior of thin-walled Zircaloy-4 fuel-clad tubes of 220 MWe Indian PHWR is studied. The chemical composition of the alloy is shown in Table 7.1. The alloy is used in the stress-relieved-annealed condition in the fuel-clad tubes. Due to the heavy deformation ($\sim 80\%$) during the prior cold-work (pilgering) process, the grains are not of uniform shape. The average grain size is of the order of 5–6 μm in the elongated direction and 2–3 μm in the other directions. From the micro-texture analysis, it was observed that the basal poles of the

Table 7.1 Chemical Compositions of Zircaloy-4

Element	Sn (%)	Fe (%)	Cr (%)	Ni (%)	O (%)	Zr (%)
wt%	1.50	0.21	0.10	0.007	0.12	Balance

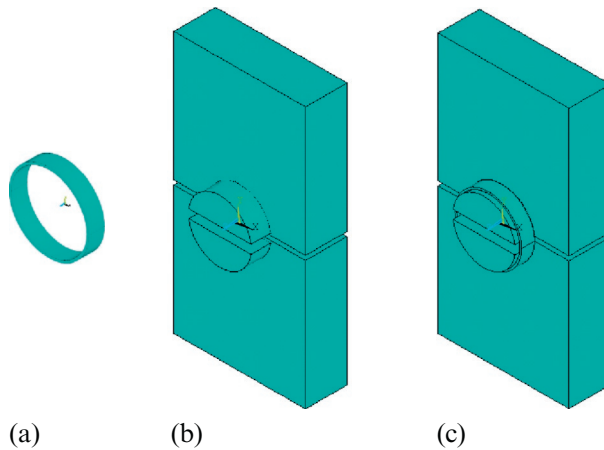
hexagonally closed packed (HCP) crystals are concentrated at about 30° from the radial direction of the tube. The micro-structure also contains Laves phases of Zr(Fe,Cr)_2 type which are considered as inclusions. After the pilgering process, stress relieving heat treatment is carried out at 480°C for 3.5 h. This relaxes the residual stresses. However, the grain micro-structure remains intact.

3 EXPERIMENT ON RING SPECIMENS FOR EVALUATION OF TRANSVERSE MECHANICAL PROPERTIES

Several specimens in the shape of rings of 3 mm width approximately were machined from the unirradiated stress-relief annealed (SRA) Zircaloy-4 fuel pins of 220 MWe Indian PHWRs. The 19-element fuel-bundle used in PHWRs is shown in [Figure 7.1](#). The inner diameter of the fuel pin specimen is 14.42 mm and the thickness is 0.4 mm. A typical ring-tension specimen is shown in [Figure 7.2a](#). The loading fixture consists of two semi-cylindrical mandrels attached to a loading rod and it is shown in [Figure 7.2b](#). The ring has been loaded on the pair of split mandrels and the test setup has been presented in [Figure 7.2c](#). The specimens have been loaded in a displacement-controlled mode until fracture and the load-displacement response of the specimens have been obtained directly from the test. The displacement data as obtained directly from the machine has been corrected for the machine compliance following the standard practice. In order to determine the transverse mechanical properties of the fuel-cladding tubes, FE analysis of the above-mentioned

**FIGURE 7.1**

19-Pin fuel-bundle used in Indian PHWRs from which ring and axially cracked fuel-clad specimens are machined.

**FIGURE 7.2**

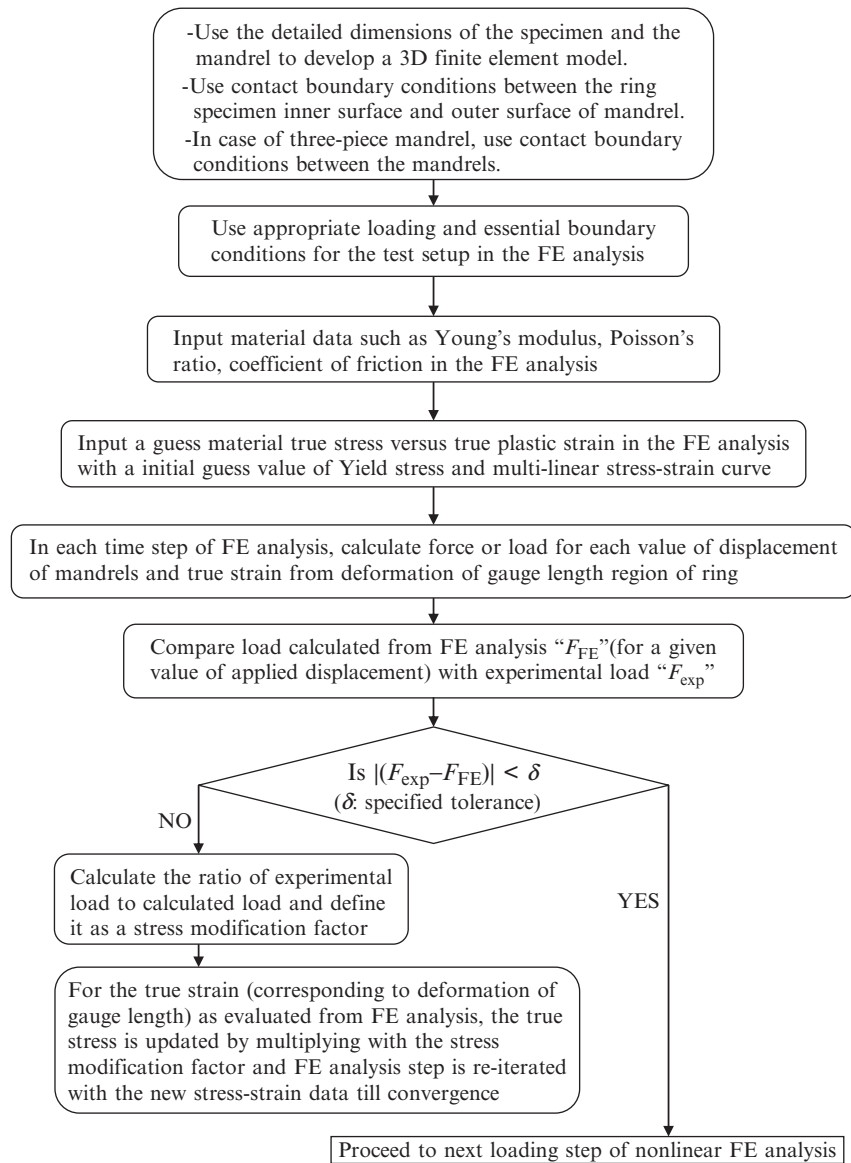
Sketch of the ring specimen, mandrel, and the ring-tension test setup.

ring-tension loading setup has been carried out and the properties have been determined using an inverse analysis procedure as discussed in the following section.

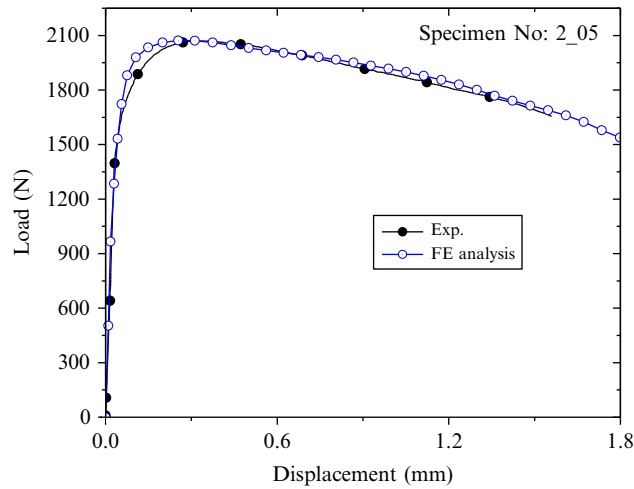
The specimen and the loading mandrel have been modeled in 3D domain. Due to symmetry, 1/8th of the test setup has been modeled with the use of suitable symmetric boundary conditions. For the nonlinear elastic-plastic FE analysis, the material properties (such as yield stress, ultimate tensile strength, and strain-hardening exponents) have been initially guessed and the load-displacement curve (as obtained from FE analysis) has been compared with that of experiment. The true stress-strain curve of the material has been modified till a satisfactory agreement of the load-displacement data between experiment and FE analysis has been obtained. This is carried out using an iterative process and the algorithm is presented in [Figure 7.3](#).

The 3D model of FE analysis takes care of effect of specimen geometry and stiffness, mandrel geometry and stiffness, contact and sliding boundary conditions between specimen mandrel, contact and sliding boundary conditions between parts of mandrel, etc., on the load-displacement behavior of the test setup with only unknown being the material true stress versus true plastic strain curve. For this purpose, an initial guess of the material stress-strain curve is used as input and an optimization algorithm is employed for calculation of actual material data. This algorithm uses the modulus of difference between the experimental and FE calculated load for a given value of applied mandrel displacement as the optimization function and updates the true stress value for the given value of true plastic strain in the gauge-length region of the specimen. These details of the algorithm employed in this work is shown in [Figure 7.3](#). After the iterative process is over, the comparison of load-displacement data between that of FE analysis and experiment has been presented in [Figure 7.4](#).

The true stress-strain as obtained through this procedure has been compared with the data obtained through the use of the conventional method in [Figure 7.5](#). In the

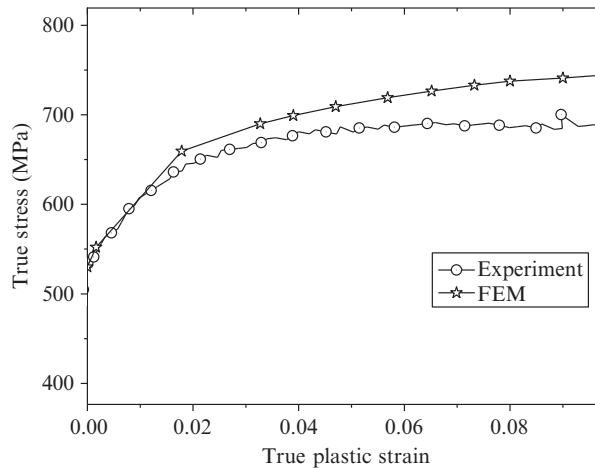
**FIGURE 7.3**

Algorithm used for evaluation of material stress-strain curve by comparing the load-displacement response obtained from ring-tension experiment and FE analysis.

**FIGURE 7.4**

Comparison of load-displacement response as obtained from FE analysis with that of experiment for a ring-tension specimen.

conventional method, the load-displacement data is converted to the engineering stress-strain data by using the information, such as, the initial area and gauge-length of the specimen. However, for this ring-tension test, the gauge-length is not known *a priori* and it has been obtained from the shape of the necked region of the specimen obtained after the

**FIGURE 7.5**

True stress-strain curve of the material Zircaloy-4 at room temperature as obtained from inverse FE analysis of the ring-tension setup and its comparison with data directly obtained from experiment.

test. The engineering stress-strain data is then converted to the true stress-strain data using standard equations of uniaxial tensile tests. The problem with this method is the determination of the gauge-length and the condition of stress-state, which is not purely uniaxial tension in this specimen (due to the curvature of the specimen). Hence, the inverse FE analysis procedure has inherent advantages for determining the mechanical properties from these nonconventional specimens as presented in Figure 7.3.

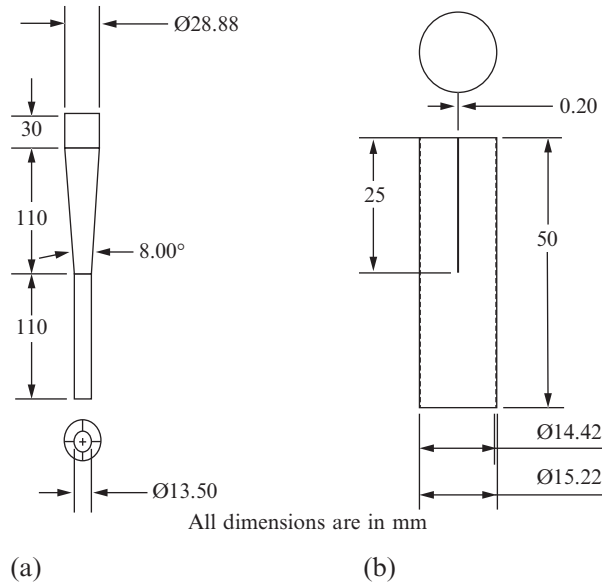
The true stress-strain curve as obtained from FE analysis has been presented in Figure 7.5 along with the stress-strain data as obtained from the conventional technique. It can be observed that the true stress-strain curve as obtained from inverse FE analysis shows a strain-hardening behavior whereas the data obtained from the conventional technique shows a flat curve after a strain of $\sim 8\%$. This material is known to exhibit substantial strain-hardening behavior at room temperature and hence, it can be observed that the data as obtained from FE analysis represents the material behavior satisfactorily.

4 FRACTURE EXPERIMENT ON TUBULAR SPECIMENS USING CONICAL MANDRELS

For evaluation of fracture resistance behavior of fuel-clad tubes, one needs to simulate the loading condition as seen by the fuel-clad inside a reactor. The fuel-pellets exert radial load on the fuel-clad due to expansion and cracking of fuel-pellets. Internal pressure also rises in the fuel-clad tubes due to accumulation of fission gas after operation of the reactor for a long time. All these loading conditions need to be simulated in the fracture tests so that the fracture behavior of the component in the actual condition can be evaluated. This data is needed for integrity analysis of fuel-clad tubes for longer operation of reactor and for achieving higher fuel burn-up. Loading with an internal conical mandrel as well as split semi-cylindrical mandrels are some of the options which have been used in this work. These tests simulate the loading condition of a fuel-clad tube in a nuclear reactor during its extended operation.

In the first fracture test setup, the conical mandrel with a half conical angle of 4° has been used. The mandrel has an increasing diameter and as it is gradually inserted into the axially cracked fuel-clad tube, the two crack-tips (on diametrically opposite sides) are loaded in mode-I. The state of stress is predominantly tensile with combination of contact pressure due to mandrel and the shear stress due to frictional interaction between the mandrel and the interior wall of the clad. Zircaloy-4 fuel-clad tubes with internal diameter of 14.42 mm, thickness of 0.4 mm, and length of 50 mm were used in the tests. The axially cracked tubular specimens have different a/W ratios ranging from 0.1 to 0.5, where a denotes the initial crack length and W denotes the total length of the specimen. The geometry and dimensions of the conical loading mandrel are shown in Figure 7.6a, whereas the details of the specimen are shown in Figure 7.6b. The mandrel has a lower straight cylindrical portion which goes inside the tubular specimen and it is used to prevent bending of the specimen during the test.

The conical portion of the mandrel opens the crack surface of the specimen when it is pushed into the fuel-clad specimen. Figure 7.7 shows the picture of the test setup with

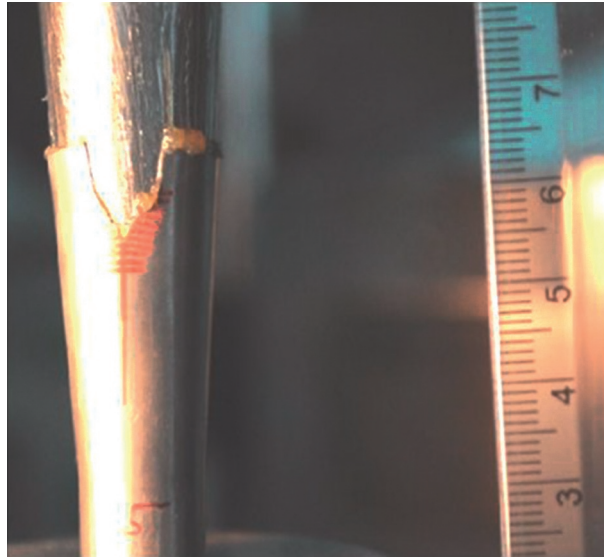
**FIGURE 7.6**

Geometry of (a) the conical loading mandrel with half angle of 4° for the conical part and (b) the axially cracked fuel-clad specimen with internal diameter 14.42 mm, thickness 0.4 mm, length 50 mm, a_0/W ratio 0.5.

initial crack-tip blunting and subsequent ductile crack growth in the specimen. A grease-based lubricant is used between the mandrel and the internal surface of the specimen as can be seen from Figure 7.7. The crack growth during loading is measured optically (through a digital camera) as can be seen from the picture. The crack-tip as well as the growth of crack front is shown with red line markings in Figure 7.7. The load and displacement data is taken from the machine directly. Using the load-displacement data (i.e., area under load-displacement curve), the total energy (elastic, plastic and frictional dissipation) used in the fracture-mechanics specimens during ductile crack growth is calculated. The effect of frictional energy dissipation is taken into account for the calculation of plastic energy dissipation required for crack growth in these alloys.

5 FRACTURE EXPERIMENT ON TUBULAR SPECIMENS USING A PAIR OF SEMI-CYLINDRICAL MANDRELS

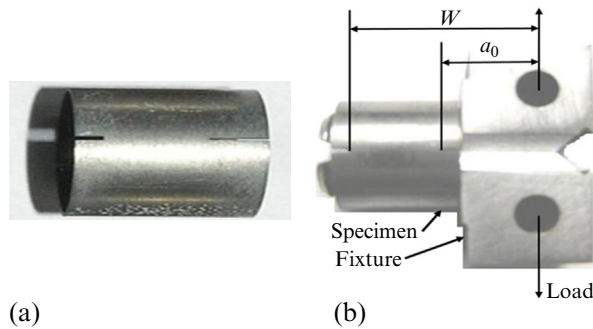
A number of axially cracked tubular specimens have been machined from the as-received fuel pins. Co-axial cracks of various sizes have been machined on these specimens and the specimens have been subjected to fatigue pre-cracking so as to create sharp initial cracks. The tubes are of 14.42 mm inner diameter, 0.4 mm thickness, and 13 mm length. A typical axially cracked fuel pin specimen is shown in Figure 7.8a. The loading fixture consists of a pair of split semi-cylindrical mandrels

**FIGURE 7.7**

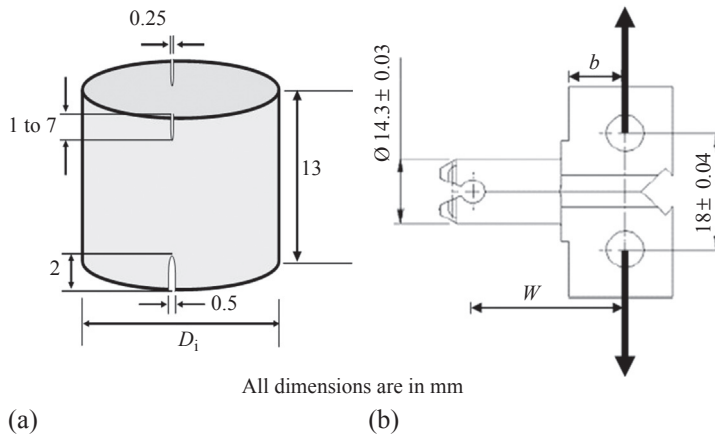
The experimental setup (axial cracked tubular specimen loaded with conical mandrel) with optical measurement of crack growth (crack-tip as well as crack advances are shown with red line markings).

which form the cylindrical holder when put together. The fixture halves are loaded through the pins after being inserted into the tubular cracked specimen. The loading arrangement is shown in Figure 7.8b. The geometrical dimensions of the specimen and the loading mandrel are shown in Figure 7.9a and b, respectively. The specimen is aligned such that the plane of axial cracks is co-planar with the openings of the split-halves of the pair of mandrels. A pin at the back end of the fixture is inserted into a groove machined in the mandrels. It serves as an anchor point about which the semi-cylindrical fixtures rotate during the loading process. The tubular specimen is prevented from slipping from the loading mandrels by inserting a clip at the rear end of the mandrels (opposite to loading grooves) as shown in Figure 7.8b.

The specimens are tested with displacement-controlled loading condition. The rate of loading is 0.01 mm/s for the test with internal conical mandrel and 0.002 mm/s for the test with pair of split semi-cylindrical mandrels. This is because the traverse of the conical mandrel is larger compared to the opening of the semi-cylindrical mandrels. The loading pins at the right end of the fixtures in Figure 7.9b are connected to the machine frames through the loading attachments. A notch as shown in Figure 7.8b is provided at the left end of the specimen in order to prevent any possible buckling during the loading. The initial crack length a_0 and width W (19 mm in the present study) of the fuel pin loading setup are labeled in Figure 7.8b. Two sets of specimens have been machined from these fuel pins. A first set of seven specimens have been

**FIGURE 7.8**

(a) An axially cracked specimen machined from fuel-clad tube; (b) specimen loaded with two semi-cylindrical mandrels to open the crack.

**FIGURE 7.9**

Dimensions of (a) an axially cracked specimen machined from fuel-clad tube ($D_i = 15.2$ mm); (b) semi-cylindrical split mandrels ($b = 8$ mm, $w = 19$ mm).

used for compliance measurements as a function of crack length to width (a/W) ratio. These compliance data has been later to used to derive the geometric or shape functions for derivation of mode-I linear elastic stress intensity factor K_I .

The initial crack length to width (a_0/W) ratios of these specimens are given as: 0.474, 0.526, 0.579, 0.632, 0.684, 0.737, and 0.789, respectively. Later, another set of six specimens with very close a_0/W ratios have been prepared for the fracture-mechanics tests. These specimens are denoted as Sp1-Sp6, respectively. The a_0/W ratios of these specimens are 0.569, 0.554, 0.595, 0.570, 0.554, and 0.570, respectively. The crack lengths during fatigue pre-cracking have been measured at nine points along the thickness and their average values have been calculated. Each specimen has been subjected to

different levels of applied displacements in order to achieve different amounts of crack growth. The initial crack lengths as well as the final crack lengths after the fracture tests have been measured on the fracture surface.

6 RESULTS AND DISCUSSION

In this section, the load-displacement results as obtained directly from the experiments with two different test setups are presented. The method to derive the fracture resistance behavior of the axially cracked fuel-lad tube in terms of J - R curve is presented. Initially, the results of test setup with conical loading mandrels are presented. This is followed by the results which have been obtained from the test setup with a pair of split semi-cylindrical mandrels. The type of axially cracked specimen remains same for both the test setup with different types of loading mandrels. However, the first test setup uses specimen with 50 and 100 mm length, whereas the second setup uses specimen with 13 mm length. The diameter and thickness of the specimens are same for both the test setups.

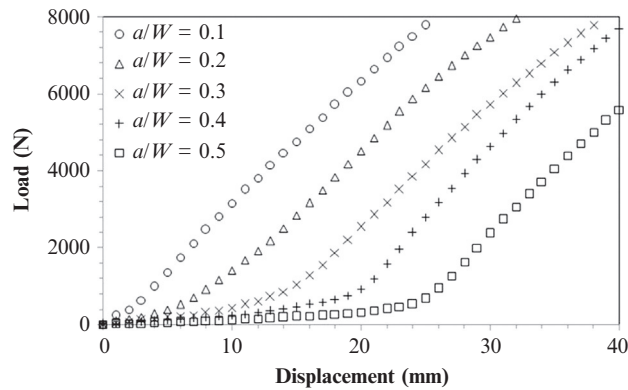
6.1 RESULTS OF THE TEST SETUP WITH CONICAL-TYPE LOADING MANDREL

The load-displacement behavior for the axially cracked specimens with different values of a_0/W is shown in Figure 7.10. As the specimen with higher a_0/W ratio has less stiffness, the load corresponding to same axial displacement of the mandrel is less. It was also observed that for the same value of applied displacement, specimens with smaller initial crack lengths absorb more energy due to extensive plastic deformation.

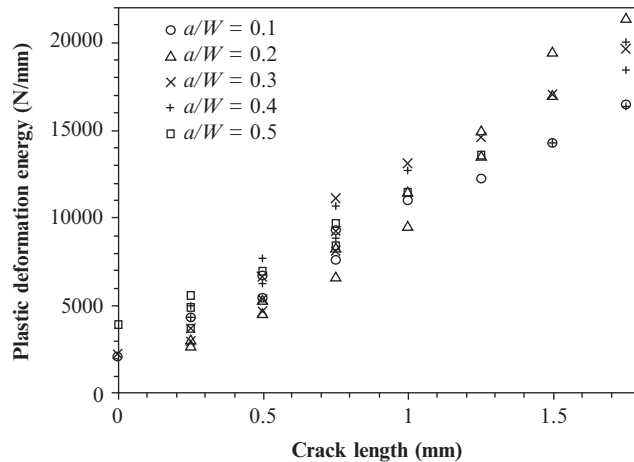
For calculation of plastic deformation energy, the frictional dissipation needs to be subtracted from the total energy applied on the specimen. Frictional dissipation is calculated from the area under the frictional force F_f versus the applied displacement curve. The force of friction can be evaluated from the equilibrium condition and is given as follows.

$$F_f = \frac{\mu F}{\sin \theta + \mu \cos \theta} \quad (7.1)$$

where μ is the coefficient of friction, θ is the half cone angle, F is the applied load. The coefficient of friction between stainless steel mandrel and Zircaloy-4 (with lubrication between the surfaces) has been taken as 0.08. The absorbed plastic deformation energy versus crack extension data was observed to be nearly independent of the a_0/W ratios of the specimens as shown in Figure 7.11. This may be attributed to the dependence of plastic deformation energy (required for ductile crack growth) on the formation of an intense localized plastic deformation zone ahead of the crack-tip, which is independent of the actual remaining ligament of the specimen. In order to compare the fracture resistance behavior of axially cracked fuel-clad specimens,

**FIGURE 7.10**

Load versus displacement behavior of axially cracked specimens with a_0/W varying from 0.1 to 0.5).

**FIGURE 7.11**

Plastic deformation energy versus crack extension data for axially cracked fuel-clad specimens with different a_0/W ratios.

the J - R curve is estimated from plastic deformation energy and elastic stress intensity factor as presented below.

The fracture resistance behavior for fuel pin specimens made of ductile materials like Zircaloy-4 is represented in terms of J - R (J -integral vs. crack extension) curves. The J -integral contains two terms, that is, elastic part J_{el} and plastic part J_{pl} , that is,

$$J = J_{el} + J_{pl} \quad (7.2)$$

The elastic part is calculated from the expression of linear elastic stress intensity factor K_I (the crack-tip is subjected to mode-I loading) for the plane stress condition as follows.

$$J_{el} = K_I^2 / E \quad (7.3)$$

where E is the Young's modulus of elasticity (90 GPa for this material at room temperature). The stress intensity factor K_I depends upon applied load P , thickness t , crack length a , width W , and geometry of the specimen and is written as

$$K_I = \frac{P}{2t\sqrt{W}} f(a/W) \quad (7.4)$$

The effect of geometry is accommodated through the term $f(a/W)$ and these are taken from Refs. [21–24] for the axially cracked thin-walled Zircaloy fuel-clad tubes. The variation of J -integral (i.e., elastic part J_{el} , plastic part J_{pl} and the sum total value of J) with crack extension for the axially cracked fuel-clad specimen with a_0/W ratio of 0.5 is shown in Figure 7.12. It can be observed that the loading process with conical mandrel involves large-scale plastic deformation as the material is highly ductile and the component is thin-walled. The extent of elastic deformation energy in the total energy required for crack growth is very limited and it is of the order of 10% as can be seen from Figure 7.12. Similar observations were made for other specimens with a_0/W ratios varying from 0.1 to 0.4, respectively.

From the load-displacement data and crack-size estimates at each data point, the plastic part of J -integral J_{pl} at each data point i is evaluated using the following expression

$$J_{pl(i)} = \left[J_{pl(i-1)} + \left(\frac{\eta_{(i-1)}}{b_{(i-1)}} \right) \frac{A_{pl(i)} - A_{pl(i-1)}}{2t} \right] \left[1 - \gamma_{(i-1)} \frac{a'_{(i)} - a'_{(i-1)}}{b_{(i-1)}} \right] \quad (7.5)$$

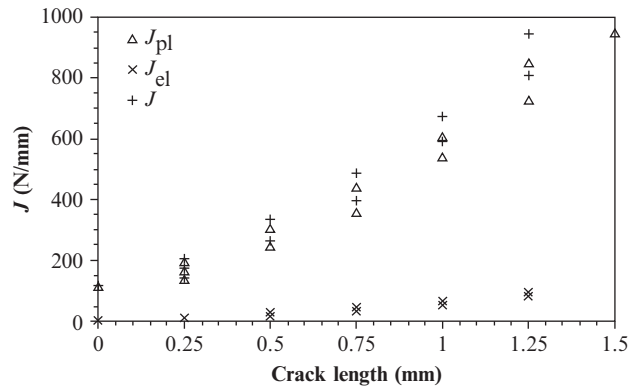


FIGURE 7.12

Variation of J -integral (elastic part J_{el} , plastic part J_{pl} , and total J) with crack extension for the axially cracked fuel-clad specimen with a_0/W ratio of 0.5.

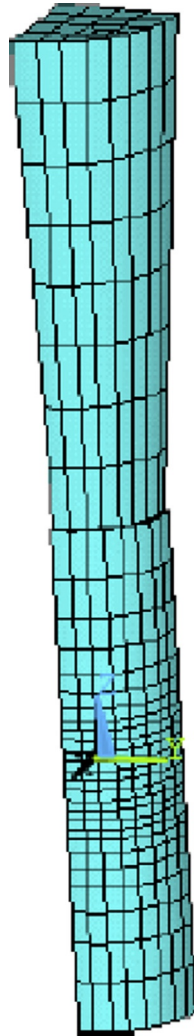
with η and γ are the geometric factor and the correction factor, respectively, for a growing crack situation. The η factor is defined as a nondimensional parameter that takes into account of the effect of geometry and loading condition at the crack-tip of a cracked specimen on the differential energy that is required for crack growth. For a growing crack, the solution for J_{pl} is corrected with a γ factor. It is defined as a non-dimensional parameter that accounts for correction in the J -integral calculation due to crack propagation, which is otherwise defined for a stationary crack.

For the axially cracked specimen loaded with conical mandrel, the state of stress ahead of crack-tip is predominantly tensile with dominant mode-I crack growth. Hence, the η and γ factors (taken from Ref. [28]) corresponding to the middle tension “M(T)” specimen has been used here for evaluation of plastic part of J -integral accordingly as described in Equation 7.5. For the case of the fuel-clad specimen with 100 mm total length (W), the value of initial crack length a_0 varies from 10 to 50 mm, respectively, and hence, the initial remaining ligament varies from 90 to 50 mm, respectively. In case of the specimen with 50 mm total length (W), the minimum value of remaining ligament is 25 mm and the maximum value is 45 mm. The crack growth measured in this work is of the order of 1.5–2 mm. The extent of crack growth and the stressed region is small compared to the actual remaining ligament and hence, the whole remaining ligament doesn’t participate in the plastic dissipation process of ductile crack growth.

The actual remaining uncracked ligament length is very large compared to the stressed region ahead of crack-tip for all the specimens with different initial crack length values. In case of this specific loading geometry (i.e., longer fuel-clad specimens loaded with conical mandrel), only a small region of the specimen, which is very close to crack-tip, will be subjected to tensile stress and hence, this region will offer resistance to crack propagation. This zone is termed here as the effective remaining ligament. This length is calculated by FE analysis and it has been found to be of the order of 20 mm. When this effective ligament length is used in the calculation of the J - R curve in Equation 7.5, the results of fracture resistance are found to be more or less similar for different values of a_0/W ratios of the axially cracked fuel-clad specimens.

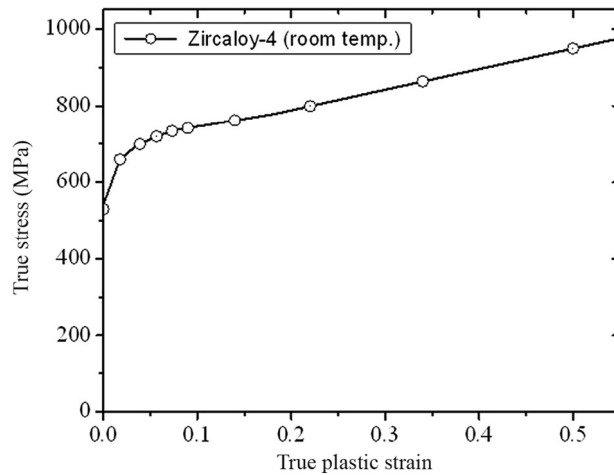
6.2 FE ANALYSIS OF THE TEST SETUP WITH CONICAL-TYPE LOADING MANDREL

FE analysis was carried out for the test setup with conical loading mandrel in order to understand the load-displacement behavior of the axially cracked specimen and to determine the effective ligament length. Due to symmetry, 1/4th of the test setup was modeled and the 3D FE mesh of the test setup is shown in Figure 7.13. For the elastic-plastic FE analysis of the test setup, the material true stress versus plastic strain data for Zircaloy-4 as shown in Figure 7.14 was used. Tests were carried out at room temperature with displacement-controlled mode and the state of von Mises equivalent stress (in MPa) and equivalent plastic strain at the crack-tip are shown in Figures 7.15 and 7.16, respectively, for an applied mandrel displacement of

**FIGURE 7.13**

3D finite-element mesh of the axially cracked specimen (with length of 50 mm and a/W value of 0.5) and the conical loading mandrel as exists in the test setup (1/4th of the test setup has been modeled due to symmetry).

30 mm. For this value of applied mandrel displacement, the von Mises equivalent stress at the crack-tip is of the order of 900 MPa (Figure 7.15) and this corresponds to equivalent plastic strain of more than 40% at the crack-tip (Figure 7.16). This intense plastic strain leads to large-scale thinning of wall of the fuel-clad tubes and ductile crack growth as can be seen from experiment (Figure 7.7). The same scale of plastic deformation at the crack-tip can be seen from the FEM simulated data as shown in Figures 7.15 and 7.16, respectively.

**FIGURE 7.14**

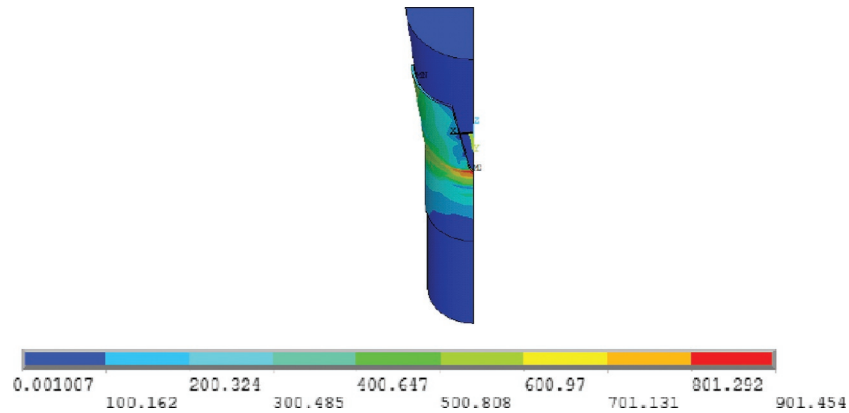
True stress versus true plastic strain of Zircaloy-4 fuel-clad tube at room temperature.

The load-displacement response of the specimen in the conical mandrel test setup (as calculated from FE analysis) is shown in Figure 7.17. This behavior is similar to the experimental data. Initially, the load is very small for an applied mandrel displacement of ~ 23 mm due to limited contact of the mandrel with the specimen crack-tip. When the crack-tip is in full contact with the conical loading mandrel, the load rises sharply and hence, there is large-scale plastic deformation at the crack-tip as can be seen from the large values of von Mises stress and plastic strain near the crack-tip in Figures 7.15 and 7.16, respectively. The extent of frictional dissipation can be computed through FE analysis from the difference of area under load-displacement curves (i.e., FE analysis with and without friction) as shown in Figure 7.18. It can be observed from Figure 7.18 that the extent of frictional in the test setup with conical mandrel is large and it is of the order of 40% of the total energy used (in the experiment) by the specimen for ductile crack growth.

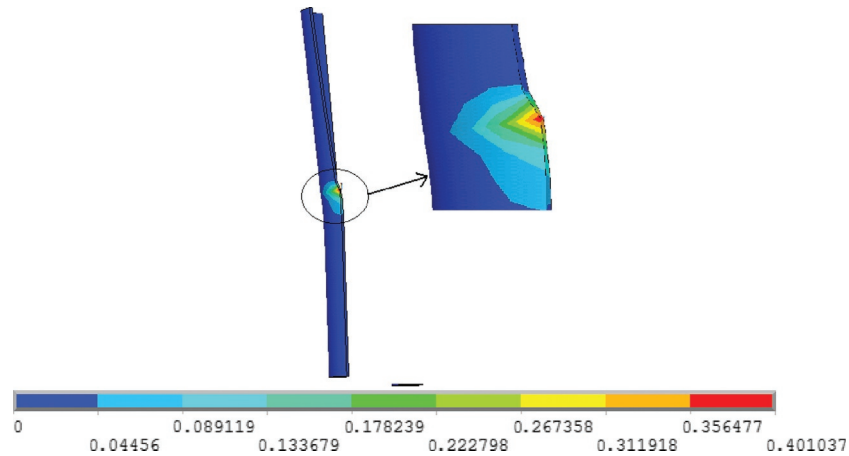
The fracture resistance behavior (in terms of J - R curve) is evaluated for the two types of axially cracked fuel-clad specimens, that is, specimens with total length of 50 and 100 mm, respectively, and these are shown in Figure 7.19. It can be observed that J - R curve is almost independent of the total length (W) of the specimen as the effective remaining ligament length is small compared to the actual remaining ligament length.

6.3 RESULTS OF TEST SETUP WITH PAIR OF SPLIT SEMI-CYLINDRICAL LOADING MANDRELS

The load-displacement curves as obtained from the tests with semi-cylindrical loading mandrels (Figure 7.8) are presented in Figure 7.20. The displacement is measured through a COD gauge mounted at the right-end opening of mandrels as shown in Figure 7.8b and the measurement of load is taken directly from the load cells. As the

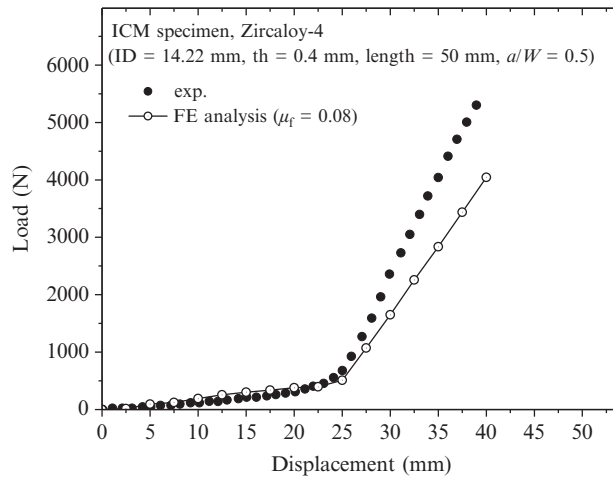
**FIGURE 7.15**

Contour of von Mises equivalent stress (MPa) in the axially cracked fuel-clad specimen (with length of 50 mm and a_0/W value of 0.5) with an applied displacement of 30 mm on the mandrel.

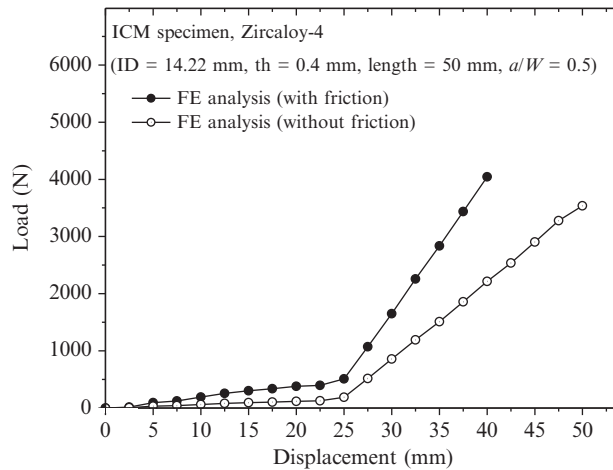
**FIGURE 7.16**

Contour of von Mises equivalent plastic strain in the axially cracked fuel-clad specimen (with length of 50 mm and a_0/W value of 0.5) for an applied displacement of 30 mm on the mandrel.

load line displacement is measured with the COD gauges, it is not necessary to correct the data for the machine compliance. The η and γ factors depend upon a/W ratios of the specimens and these are taken from Refs. [21–24]. The load-normalization technique has been used to estimate the crack extension during loading of the cracked fuel pin specimens.

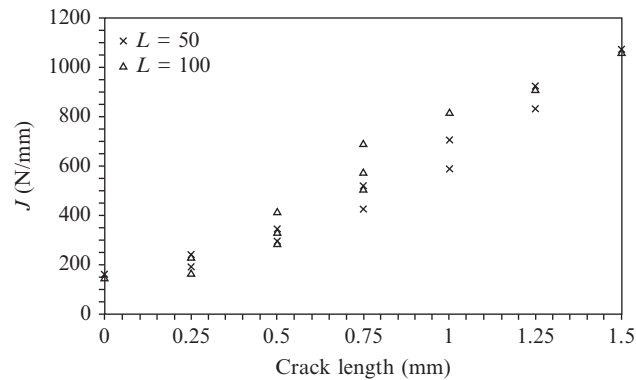
**FIGURE 7.17**

Load-displacement response of the fuel-clad specimen (with length of 50 mm and a_0/W value of 0.5) in the conical mandrel test setup (exp. vs. 3D finite-element analysis).

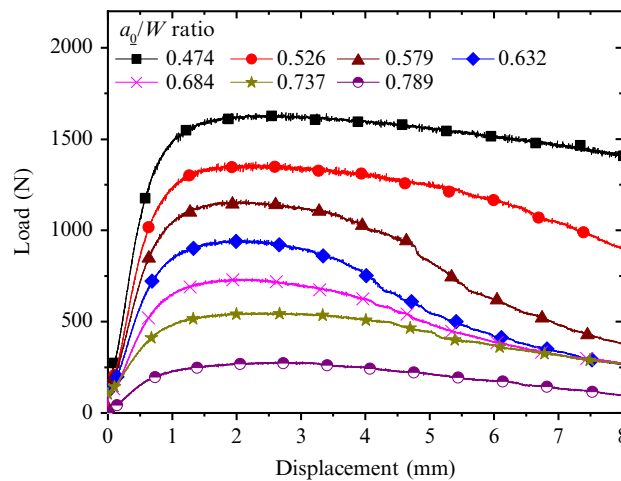
**FIGURE 7.18**

Effect of friction on load-displacement response of the fuel-clad specimen (with length of 50 mm and a_0/W value of 0.5) in the conical mandrel test setup.

The load-normalization method was at first proposed by Herrera and Landes [29]. According to this method, the load can be normalized by a geometry function (i.e., the load is assumed as a product of a pure geometric function and a pure deformation function). The resulting function, known as the pure deformation function, can be expressed through a relationship between the normalized load and the normalized

**FIGURE 7.19**

Fracture resistance behavior (in terms of J - R curve) of the axially cracked fuel-clad specimen (with a_0/W value of 0.5) for two different values of total length (i.e., 50 and 100 mm).

**FIGURE 7.20**

Load-displacement curves of fuel pin specimens with different a_0/W ratios when loaded with two semi-cylindrical mandrels.

plastic displacement. To determine the instantaneous crack growth during loading, the deformation function can be described in terms of a power law form [30] with two calibration points or by LMN function [31] with three calibration points. The calibration points are the data points at which the load, displacement and crack length must be known simultaneously. These data points are usually determined from the fracture surfaces of tested specimens.

For evaluation of the fracture resistance behavior of the axially cracked fuel pin specimens, additional set of six specimens which have close a_0/W ratios (i.e.,

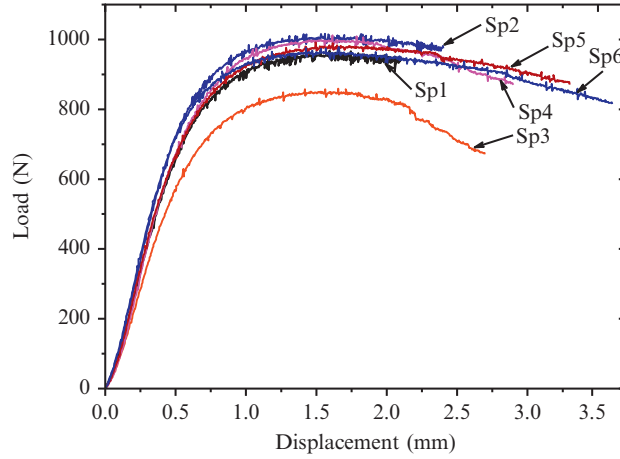


FIGURE 7.21

Load-displacement curves of the six different specimens as obtained from fracture toughness tests.

0.554–0.595) have been used. The load-displacement curves as obtained from the fracture tests are presented in Figure 7.21. Each load-displacement curve has been normalized according to the following procedures. The load P_i at any point i , up to the maximum load is converted to a normalized load P_{Ni} as [32–34]

$$P_{Ni} = \frac{P_i}{2tW[(W - a'_{bi})/W]^\eta} \quad (7.6)$$

The term a'_{bi} is the blunting-corrected crack size at the i th data point and is written as

$$a'_{bi} = a'_0 + J_i/2\sigma_Y \quad (7.7)$$

where a'_0 is the crack length at the start of the test and σ_Y is the material flow stress (i.e., average of the yield and ultimate stress), J_i is the value of J -integral corresponding to point i in the load-displacement diagram. The J -integral value J_i is calculated using Equations 7.2–7.5, respectively. J -integral represents the product of material flow stress σ_Y , a constant m (taken as 2), and crack growth Δa . Hence, to have an estimate of crack growth Δa , the J -integral corresponding to the load step i is divided by $2\sigma_Y$ and is added to the initial crack length a_0 in order to calculate the current crack length. The updated crack length is used for further calculation of normalized load and plastic displacement according to Equations 7.6 and 7.8, respectively. The normalized plastic displacement v_{pli} at point i is evaluated as

$$v'_{pli} = (v_i - P_i C_i)/W \quad (7.8)$$

where C_i is the elastic compliance of the specimen based on the blunting-corrected crack size a'_{bi} and it is calculated using the following equation for the fuel pin specimens [23].

$$C(a/W) = -0.01752 + 0.10473(a/W) - 0.20424(a/W)^2 + 0.13573(a/W)^3 \quad (7.9)$$

The final load-displacement pair corresponding to the end of loading is also normalized using the same equations as described above except that the final measured crack size a'_f (after the specimen is break-open) is used instead of the crack size evaluated using the normalization curve.

The normalized load-displacement data for all the six specimens have been obtained using the above-mentioned approach and these are presented in Figure 7.22. The data points to the right-hand side of the anchor point are excluded as these data points corresponds to that of crack propagation and the crack growth data at each point needs to be subsequently evaluated. The data with $v'_{pl} \leq 0.001$ are also discarded due to absence of crack growth at this very small amount of plastic displacement. The remaining data points are fitted with the following analytical normalization function, that is,

$$P_N = \frac{a + bv'_{pl} + cv'^2_{pl}}{d + v'_{pl}} \quad (7.10)$$

where a , b , c , and d are fitting constants. Using these values of the constants, the crack growth information of the remaining data points are evaluated.

The process of estimation of crack growth using load-normalization technique involves an iterative procedure. The crack growth at a point is guessed and the normalized load and displacement data are recalculated using Equations 7.6 and 7.8, respectively. The procedure is repeated with update in the crack growth information till the data point falls on the curve defined by Equation 7.10 within a predefined tolerance. In this way, the crack extensions at all the points of the load-displacement diagram are estimated. Using this information of crack growth as a function of loading, the

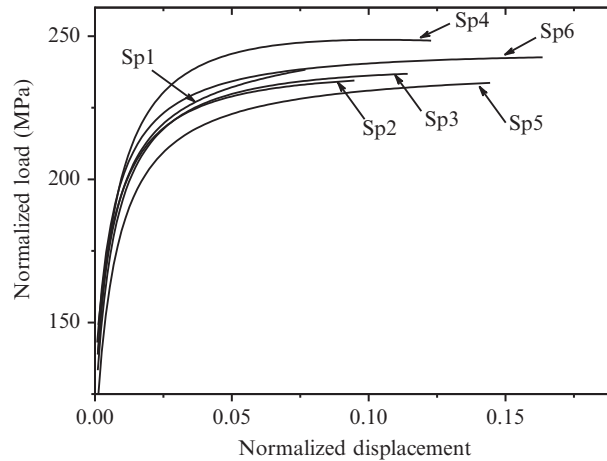
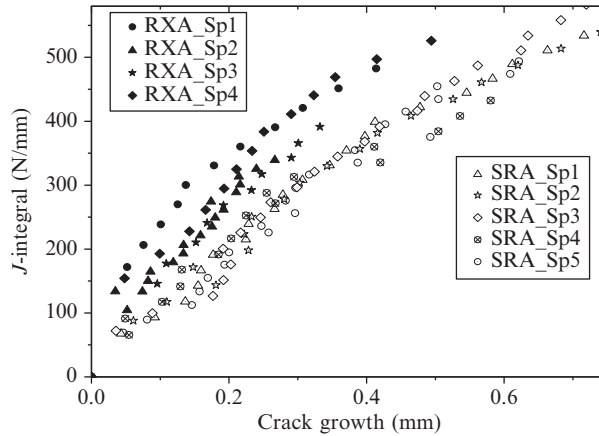


FIGURE 7.22

Fitted curves for normalized load-displacement data for all the six specimens used in the fracture test.

**FIGURE 7.23**

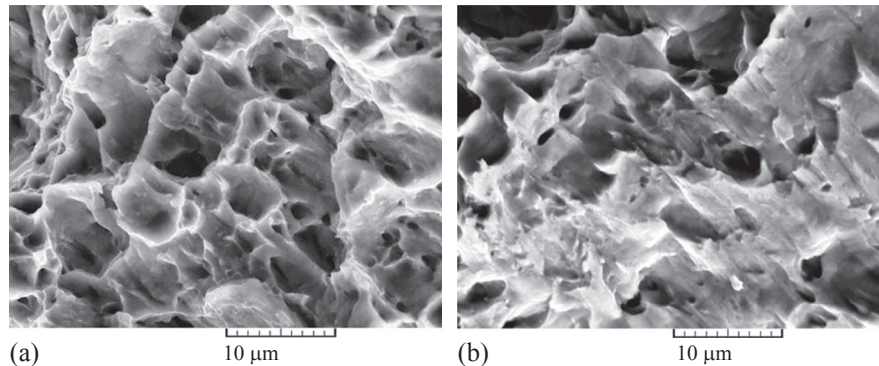
Fracture resistance behavior of axially cracked fuel-clad specimens in terms of J - R curve for RXA- and SRA-type zirconium alloys.

fracture resistance of the specimens in terms of J - R curve is evaluated and these are presented in Figure 7.23 for recrystallization-annealed and SRA types of zirconium alloys.

The J - R curves of Zircaloy-4 fuel-clad specimens as obtained from both the test setups can be compared, that is, Figure 7.19 versus Figure 7.23. In the loading setup with a pair of split semi-cylindrical mandrels, crack growth values of the order of 0.7 mm were obtained as the remaining ligament is small. However, for the loading setup with conical mandrel, crack growth of the order of 1.5 mm has been obtained as the actual remaining ligaments were of sufficient length. For a given extent of ductile crack growth (e.g., 0.6 mm), the average J -integral values are of the order of 500 N/mm for both the test setups as can be seen from Figures 7.19 and 7.23, respectively.

The SEM images of fracture surfaces of axially cracked Zircaloy-4 fuel-clad specimens with a_0/W values of 0.474 and 0.632 are shown in Figure 7.24a and b, respectively. The fracture surface shows dimpled features representing void nucleation, growth, and coalescence phenomena for two specimens with different a_0/W ratios. There is adequate shear deformation in the ligament of the adjacent voids which shows the final necking of the voids leading to stable crack growth. This shearing process is more dominant in Figure 7.24b compared to Figure 7.24a and this can be attributed to the higher a_0/W ratio of the specimen. The remaining ligament is smaller in specimens with higher a_0/W ratios and it can lead to shear-dominated stable crack growth in addition to void nucleation and growth process which leads to dimpled fracture surface.

Hence, it may be concluded that with proper accounting of the frictional energy dissipation and plastic deformation energy in case of the test setup with conical mandrel, it is possible to obtain accurate values of fracture resistance behavior of the axially cracked fuel-clad specimens. Both the test setups (with different types of mandrels) provide similar fracture resistance data owing to nearly similar state of stress at the crack-tip.

**FIGURE 7.24**

SEM images of fracture surfaces of axially cracked Zircaloy-4 fuel-clad specimens with a_0/W value of (a) 0.474 and (b) 0.632.

7 CONCLUSION

It is important to ensure the integrity of the thin-walled Zircaloy fuel-clad tubes in a nuclear reactor in order to prevent the release of radioactive fission products to the reactor environment.

For longer residence time in the reactor, these tubes should have superior mechanical properties and adequate resistance to axial crack propagation. In this work, the ring-tension tests were carried out on ring-specimens machined from the fuel-clad tubes. Both the specimen and the loading mandrel were modeled through FE analysis in order to evaluate the load-displacement behavior of the test setup. By comparing the load-displacement response as obtained from FE analysis with that of experiment, the material stress-strain curve was evaluated following an inverse analysis procedure.

The fracture resistance behavior of axially cracked tubular specimens machined from the same fuel-clad tubes has also been evaluated. Specimens with different a_0/W ratios ranging from 0.1 to 0.5 were tested in two different test setups. Two different geometries of loading mandrels were used in order to vary the state of crack-tip stress in the axially cracked tubular components. The total plastic deformation energy versus ductile crack extension data was observed to be more or less same for different values of initial crack lengths of the specimens in the test setup with conical mandrel. This indicates that the ductile fracture resistance behaviors (in terms of J - R curve) of long thin-walled axially cracked tubes are nearly independent of initial crack lengths provided the crack-tip is far away from the boundaries. This method can be applied to the tubular components of different industries in order to evaluate their fracture resistance properties. This will help in integrity analysis and residual life assessment of the critical thin-walled tubular components.

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Seismic risk of RC water storage elevated tanks: Case study

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1 INTRODUCTION

The phenomenon of earthquakes is not new or specific to Algeria. It appears cyclically from one region to another causing sometimes immeasurable damage that no other natural forces can cause. In the most recent, catastrophic earthquakes (see [Table 8.1](#)) demonstrate the weakness of civil engineering structures relative to the destructive power of these events [1].

Given that seismic prediction is still far from becoming a reality, it is necessary to improve the seismic behavior of existing structures. Seismic design mainly focuses on ensuring an acceptable level of safety to reduce risks related to failures, disasters, and loss of life. The need to improve the methodology of structures calculation subjected to seismic action has been widely found in various seismic codes. In seismic codes, the study of the response of structures under seismic action is conducted by using methods depending on the nature of the structure and its intended end use.

Given the social importance of tanks and the role they play in the daily life of a country, they are sensitive structures to public opinion: the closure of a tank, even temporarily, for maintenance or repair causes strong reactions from subscribers, quick to denounce the laxity or negligence of public services. This is the reason why studies of seismic vulnerability of structures have been developed to evaluate the expected damage in the different types of structures when there is a severe earthquake [2].

This chapter focuses on the large-scale assessment of the seismic vulnerability of existing water storage elevated tanks. The problem is approached on a methodological term and illustrated by a practical example. [Section 2](#) of this chapter focuses on seismic vulnerability of structures. By reviewing some methods and approaches used in the diagnosis of vulnerability to earthquakes, we aim to provide a wide overview of the important parameters in the seismic tank behavior. The aim is both educational and operational. [Section 3](#) describes the Italian GNDT (Gruppo Nazionale per le Difese dai Terremoti) method and their different applications to structures and urban centers. [Section 4](#) contains a general presentation of the assessment method of the seismic vulnerability index. This method, based on the calculation of a vulnerability index (I_V), allows the classification of the examined water storage elevated tanks in three levels of seismic behavior. [Section 5](#) is dedicated to a practical example of a

Table 8.1 Some Previous Earthquakes

Epicerter	Country	Year	Magnitude
Kobe	Japan	1995	7.3
Izmit	Turkey	1999	7.6
Boumerdes	Algeria	2003	6.7
Kashmir	Pakistan	2005	7.6
Sichuan	China	2008	7.9
Haiti	Haiti	2010	7.0

water storage elevated tank in order to test the reliability of the developed method. The tank chosen is built on piles in Tizi Ouzou municipality, classified as average seismicity area. Finally, the main conclusions and the lessons learnt from the experience are given in [Section 6](#). A fairly rich bibliography is given at the end of the chapter.

2 SEISMIC VULNERABILITY

The seismic vulnerability of a structure is a quantity associated with its weakness in the case of earthquakes of given intensity, so that the value of this quantity and the knowledge of seismic hazard allows us to evaluate the expected damage from future earthquakes [3]. The evolution of the vulnerability study in the world gave birth to several methods of assessing damage to the individual buildings, structures, buildings, urban and historical centers. They differ from one another in terms of cost, in used facilities, and in precision. But there is, to our knowledge, no method of seismic evaluation of water tanks.

The systemization of these vulnerability assessment approaches has been developed by many researchers and therefore differs due to varying levels of dependence of the following factors: nature and objective of the assessment, quality and availability of information, characteristics of the structure stock inspected, scale of assessment, methodology criteria, and degree of reliability of the expected results. Because of these differences, the coherence and consensus regarding the classification is always a contentious issue. Various methods for seismic risk assessment have been implemented in countries with strong seismicity; they were ranked in ascending order of complexity, such as:

The simplified method developed by Zacek [4] for use by architects written in 1993, is a rapid and inexpensive method for assessing a presumption of buildings vulnerability to earthquakes by visual inspection. It can be used by nonexperts to identify potentially dangerous buildings. The review duration is about 30 min by construction. This method, presented as a guide, was used in Pointe-à-Pître (Guadeloupe) and Fort-de-France (Martinique) to assess the vulnerability of buildings.

A second method is the one proposed by the working group “vulnerability of existing buildings” of the French Association for Earthquake Engineering (AFPS) [19]. This method is applicable to buildings constructed since 1960. Two levels of assessment are covered. The first is a qualitative approach and the second in a quantitative verification. For the qualitative approach, an evaluation grid is filled in which a penalty coefficient is assigned to the different vulnerability factors. These coefficients are used to calculate a coefficient K varying from 0 to 100. Depending on its value, it is concluded of a presumption of structure vulnerability, ranging from high vulnerability to very low vulnerability, considering intermediate levels.

For the ATC 40 method [20], designed for reinforced concrete constructions, published in the United States in 1986 by Applied Technology Council, is innovative compared to conventional approaches. Vulnerability is assessed by a nonlinear static

analysis. Displacement capacity curves (therefore deformability) are plotted for the buildings studied. These curves depend on the characteristics of these buildings and not a seismic load. The different degrees of damage corresponding to the displacements on the curve are located. By correlating the capacity curve of the displacement building with a maximum displacement caused by seismic motion of the soil given, determined by a proposed method in the document, a point called “performance point” is obtained. Its position, relative to the desired level of performance, indicates that this level is reached or not.

The Federal Emergency Management Agency (FEMA) 310 method developed in 1998 and published in the United States by the FEMA is presented as a manual [21]. In this method, three levels of evaluation are proposed, level 1 is a presumption of vulnerability, level 2 is quantitative analysis, and level 3 is in-depth analysis.

The HAZUS 99 method developed in 1999 and edited in the United States by the FEMA is presented as a manual. This manual provides a detailed methodology for assessing vulnerability to urban scale. The aim is to provide an estimate of losses that area or region may suffer during an earthquake. In this method, three types of analysis are proposed (briefly, standard, and deepened).

In Italy, the procedure to qualify the seismic vulnerability of buildings has been proposed for the first time by Benedetti et al. [5] and then this method was developed by the Gruppo Nazionale per la Difesa dai Terremoti of the Italian Consiglio Nazionale delle Ricerche (CNR-GNDT). This method which is the subject of our interest will be explained in [Section 3](#).

We can also mention the RISK-UE project [22] which is an advanced approach to earthquakes risk scenarios with applications to different European towns, financed by the European Commission which is a major step toward the development of a European methodology to assess and mitigate the seismic risk of lifelines and essential facilities in Europe. The RISK-UE project is based on the assessment of the vulnerability index for a given current building. This index is a function of the constructive typology and different factors that can change its behavior. The RISK-UE project was used to treat the seismic vulnerability of seven pilot European cities (Thessaloniki, Catania, Nice, Bucharest, Sofia, Barcelona, and Bitola) [6].

3 VULNERABILITY METHOD OF CNR-GNDT

This method evaluates the seismic vulnerability of buildings, determining as a normalized index of vulnerability, which is obtained with the aid of survey forms which are filled in for one building or a group. The CNR-GNDT method was originally applied to stone rubblework buildings and reinforced concrete buildings, and a very detailed procedure exists for each one. The method is very easy to apply and it considers the structural characteristics of the buildings, such as type of construction, building's use, quality of materials, structural system, geometric aspects of the structure, building's state of preservation, etc. These characteristics are quantified as parameters and are evaluated considering 11 factors, and a class, score, and weight

factor is assigned to each one. There are four classes for stone masonry buildings: A, B, C, and D; and three for reinforced concrete buildings A, B, and C.

Several studies have been dedicated to the CNR-GNDT method, in some countries in South America, Europe, and North Africa, such as:

Gent Franch et al. [2], in their study, the Italian CNR-GNDT vulnerability index for masonry buildings was modified in order to obtain a vulnerability index I_{3GA} that considers the properties of confined masonry shear-wall buildings. In Chile, the modified Italian method was applied considering a sample of 24 confined masonry buildings with three and four storeys built during the last 25 years for social housing programs.

Vicente et al. [7] applied methodology considered a hybrid technique in the evaluation of the vulnerability of the entire historic city center of Coimbra (Portugal). The vulnerability index formulation proposed is based essentially on the GNDT approach, for the vulnerability assessment of residential masonry buildings. This methodology is based on post-seismic damage observation and survey data covering a vast area, focusing on the most important parameters affecting building damage which must be surveyed individually. Overall vulnerability is calculated as the weighted sum of 14 parameters used in the formulation of the seismic vulnerability index.

Bezzazi et al. [8] have adapted a scale of vulnerability from the CNR-GNDT method, where a retiming of the weighting coefficients is performed, to better match with architectural characteristics of local buildings in northern Morocco. The authors have even proposed fragility curves for this typology building.

A large-scale vulnerability assessment using two methods was conducted for buildings of Tunis by Mansour et al. [9], for which few data were available. Vulnerability methods inspired by the EMS98 concepts and the Italian GNDT concepts were modified and applied to pilot-scale buildings located in the downtown zone (Habib Bourguiba Avenue) and in the old zone (Medina).

4 GENERAL PRESENTATION OF THE VULNERABILITY INDEX METHOD

An accurate assessment of the tank vulnerability to earthquakes is long, laborious, and expensive, it is better to perform an initial study which can be only qualitative, to determine if a more thorough approach is needed, which also allows to thinning down the problem. This approach, allows evaluating a presumption of vulnerability (that is to say, a pre-diagnosis) that can be done by an engineer trained in seismic construction, without assistance of an experienced engineer of design office in engineering. It includes the study of available archival material, if possible a search for information from individuals who participated in the construction of the project, a statement of vulnerability factors in an external and internal visual inspection on site, and operation of grids field surveys, covering architectural design, bracing system, and the site.

This method is a simplified empirical method inspired by the CNR-GNDT method. Its goal is to allow for the establishment of a seismic inventory of water storage tanks on a large scale by the simple and rapid recording of structural parameters. We have made some modifications to the methodology developed by Benedetti et al. [5], where a resetting of the weighting coefficients is performed in order to better fit to the geometric characteristics of elevated tanks. The main differences are the introduction of three new parameters called clashing, seismic zone, and location of the reservoir. Readers wishing for further details about this method can consult the reference by Hammoum et al. [10].

In Table 8.2 [11], for each parameter, we define three classes of vulnerability. And for each class, will correspond an elementary score K_i . Each parameter is affected by a weighting factor W_i . The elementary score K_i of each parameter can take three values (0, 5, or 15). The value of the weightings W_i varies from 0.25 to 1.75. The value of 0.25 corresponds to a minimum penalty of the parameter and 1.75 for a maximum penalty considering intermediate situations, as shown in Table 8.2. The partial score of a parameter is obtained by the product ($K_i \cdot W_i$) and the vulnerability index I_V is expressed as the sum of the partial scores of different parameters:

$$I_V = \sum_{i=1}^{12} K_i \cdot W_i \quad (8.1)$$

Considering all the 12 analysis criteria, listed above, we propose the following classification, divided into three levels of vulnerability (Table 8.3).

Table 8.2 Classification of Containers According to Their Vulnerability Index I_{V5} [11]

No	Parameters	Elementary Score K_i			Weighting Factor W_i
		Class A	Class B	Class C	
1	Seismic behavior	0	5	15	1.75
2	The clashing	0	5	15	1.75
3	The bracing system	0	5	15	1.50
4	Regularity in plan	0	5	15	1.50
5	Regularity in elevation	0	5	15	1.50
6	Soil type	0	5	15	1.00
7	Site effect	0	5	15	1.00
8	Seismic zone	0	5	15	1.00
9	Quality of the nodes	0	5	15	0.75
10	Qualities of the bracing system	0	5	15	0.75
11	Maintenance	0	5	15	0.50
12	Details	0	5	15	0.25

Table 8.3 Classification of Tanks According to Their Vulnerability Index I_V [11]

Level	Green	Orange 1	Orange 2	Red
I_V	0-33	33-90	90-133	133-199

- The green level, $0 \leq I_V \leq 33$: The examined tank is not vulnerable. It presents a good behavior to the earthquake and it doesn't require special repairs.
- The orange level 1: $33 \leq I_V \leq 90$: The tank has an average vulnerability to earthquake. It needs monitoring.
- The orange level 2: $90 \leq I_V \leq 133$: The tank has fairly high earthquake vulnerability.
- The red level: $133 \leq I_V \leq 199$: The tank has high earthquake vulnerability. Therefore, the tank must be decommissioned immediately.

5 CASE STUDY

As a practical application, we illustrate the method of vulnerability index in a water storage elevated tank of 100 m^3 capacity and located in average seismicity zone in Algeria. This tank shown in [Figure 8.1](#) of frustoconical form has been made of reinforced concrete in 1986 and has been investigated and analyzed 28 years later. Several visits were made to the archives of the Hydraulic Administration of Tizi Ouzou. We were able to find the formwork drawings of our tank dating from 1986. No note of calculation has been found. A detailed review of these execution plans (see [Figure 8.2](#)) found in very good condition after 28 years of archiving have enabled us to evaluate the seismic load in [Section 5.3](#). The geometrical characteristics of the tank are given in [Table 8.4](#).

5.1 THE BRACING SYSTEM

The bracing system of an elevated tank is constituted by the set of resistive elements ensuring its rigidity and stability to the horizontal forces caused by the earthquake. It constitutes one of the most important aspects of seismic design. The classification of this parameter is defined taking into account the capacity of energy dissipation to the seismic action and deformation. For this, there is currently interest to equip tanks with a regular structural mesh in order to avoid twisting of bracing elements. The tank studied is installed on a supporting system of pilings which is a structure that has a low rigidity distribution of mass and stiffness that changes from abrupt manner (see [Figures 8.1](#) and [8.2](#)). The parameter is classified as **class C**.

5.2 QUALITIES OF THE BRACING SYSTEM

For this parameter, classification is done according to the quality control of materials and quality control of execution. Systematic tests on the materials used must be made by the company on site during construction, in order to control the concrete strength. This task

**FIGURE 8.1**

The general view of Sikh Oumedour tank.

should include supervision of conducted tests on materials, in order to meet general terms and conditions of design, calculation, and execution. These quality controls are used to limit the problems of form stability, the ductility, and the materials capacity to deform plastically during high loads without significant reduction of strength.

During the consultation of the file in the archives, we are interested in the quality of the concrete tank. We could read that the tests were conducted at a rate of three samples of three compression test pieces of each, respectively, for 7, 14, and 28 days for each corresponding element resistant to a class of concrete (foundations, supporting system, vessel, the dome cover, etc.). From the file, the concrete used for the different resistant elements have a characteristic strength of concrete between 20 and 25 MPa. These concretes are of average quality because they were carried out on site in conditions far removed from those of concrete mixing plants equipped with measurement and weighing instruments. The parameter is classified as **class B**.

5.3 SEISMIC BEHAVIOR

The seismic action results in an abrupt horizontal or vertical displacement of the soil, causing the deformation of the foundations and underground parts of the tank. In a partially filled tank, seismic excitation puts part of the fluid in motion, which leads to

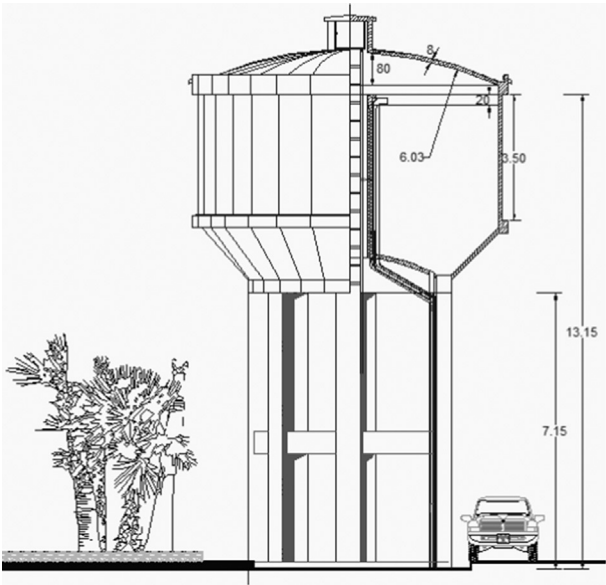


FIGURE 8.2
Longitudinal cross-section of the tank.

Table 8.4 Geometrical Characteristics of the Tank

Theoretical capacity of elevated tank	100.00 m ³
Internal diameter of the tank D	6.00 m
External diameter of the chimney	1.20 m
Wet wall height above the intermediate belt	3.15 m
Height above the free surface of the liquid	0.20 m
Total height of the wall between the two belts	3.35 m
Thickness of the dome	0.08 m
The flitch of the dome f	0.80 m
The radius of curvature of the dome R	6.03 m
The area of the dome S	30.27 m ²
<i>Real capacity of the tank</i>	104.40 m ³

the formation of surface waves. This phenomenon is called the hydrodynamic effect, and its appreciation can be given by the Housner method [12], which is based on the decomposition of the water action in impulse and oscillation action. Under the seismic action effect, the overall stability of the tank to collapse must be satisfied. The overall stability relating to rigid body behavior can be lost by sliding or overturning. For this purpose, it is necessary that the stabilizing moment be greater than the overturning moment. The classification of this parameter is done according to a safety factor “ F_s ” defined as the ratio of the stabilizing moment and overturning moment.

5.3.1 Hydrodynamic effect assessment by the approximate method of Housner

It is futile to try to protect a structure against the effects of earthquakes simply by virtue of the calculations said antiseismic. Experience has shown that structures whose design was correct and the rules correctly applied have a fairly high probability to support adequately a seismic action of moderate intensity. The choice of the calculation method and the structure modeling must have the objective that best reproduces the real behavior of the structure. During the seismic design of water storage elevated tank with a capacity more or less important, it may be a matter of studying the hydrodynamic effect of water on the structure in accordance with the Algerian Seismic Code [13] when the capacity is equal or greater than 1500 m³ in seismic zones (II and III). The evaluation of hydrodynamic forces is a crucial step. For this reason, we use the analytical method of Housner to estimate the response of a liquid in rigid tanks (cylindrical or rectangular), seismically excited. In the case of a water storage elevated tank, we cannot consider the vessel as being rigidly related to the soil and therefore undergoing the same value of the maximum acceleration that the soil. Indeed when the vessel is on top of a structure, we must consider the flexibility of the latter. Approached calculation by the Housner method consists of modeling all the water storage elevated tank represented by its equivalent mechanical and mathematical model (Figure 8.3).

The action of the liquid is divided into two actions: A passive action causing pulse actions and active action causing oscillation actions [14]. The pulse efforts come from that a portion of the liquid mass, called passive mass, reacts by inertia to the translation of a tank walls. Its equivalent mechanical system is obtained by considering a mass M_i , rigidly connected to the tank at a height h_i as it applies on the walls, the same horizontal forces that the equivalent water mass. As for oscillation efforts, they come from that another part of the liquid mass, called active mass is set in oscillation motion under the earthquake action.

In summary, the total water mass M_e can be divided into a passive mass M_i and an active mass M_o rigidly connected on the one hand and by means of a spring constant K_1 on the other hand.

In the adopted mathematical model for a water storage elevated tank (Figure 8.3), the mass M_o is connected to the structure by a rod of the same stiffness K_1 ensuring direct coupling with M_1 , while M_i is connected to the ground by a rod representing

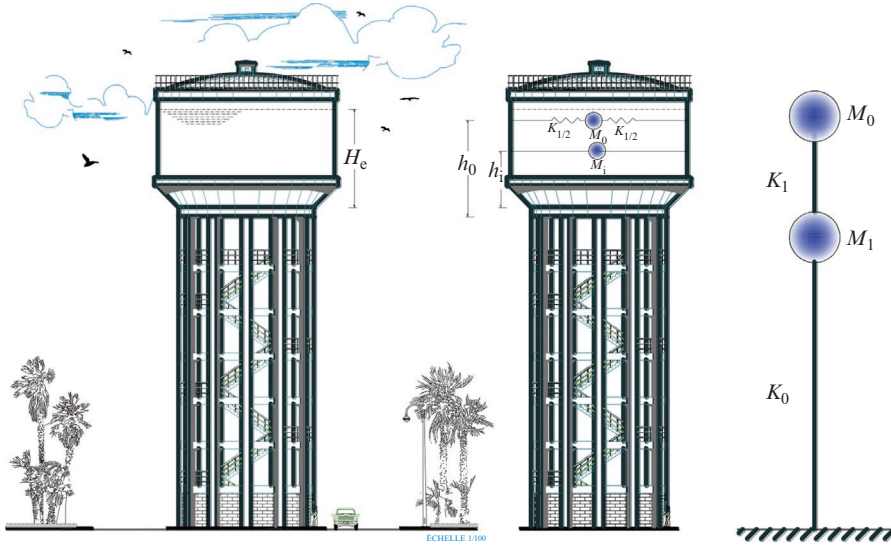


FIGURE 8.3

The elevated tank, the mechanical equivalent schema, and the mathematical model adopted.

the supporting system of the structure of spring constant K_0 . The system has two degrees of freedom.

5.3.2 Evaluation of the masses

The values of M_i , M_0 , h_i , h_0 depend only on the geometry of the tank and may be calculated by the relations developed by Housner.

Mass M_1 is given by the following formula:

$$M_1 = M_i + M_c + \frac{33}{140} \cdot M_{\text{tower}} \quad (8.2)$$

Where the inert mass (passive) M_i is given by the following formula:

$$M_i = M_e \cdot \frac{\text{th}\left(\frac{R}{H_e} \sqrt{3}\right)}{\frac{R}{H_e} \sqrt{3}} \quad (8.3)$$

We indicate by R and H_e , respectively, the interior radius of the vessel and the height of water in the tank. By M_c and M_{tower} , respectively, the vessel mass and the supporting system mass.

The oscillation mass (active) M_0 is given by the following formula:

$$M_0 = M_e \cdot 0.318 \cdot \frac{R}{H_e} \text{th}\left(1.84 \cdot \frac{H_e}{R}\right) \quad (8.4)$$

The calculation of the different masses is given in the [Table 8.5](#).

Table 8.5 Evaluation of Different Masses

Mass of the water M_e	104,404.08 kg
Mass of the vessel M_c	44,327.83 kg
Total mass M_{tot}	148,731.91 kg
Mass of the supporting system M_t	12,870.00 kg
Inert mass M_i	79,704.25 kg
M_1	124,032.08 kg
M_0	22,677.45 kg

5.3.3 Evaluation of stiffnesses

The fundamental pulsation of the water vibration in the vessel is given by:

$$\omega_0^2 = \frac{g}{R} \cdot \sqrt{\frac{27}{8}} \cdot \text{th} \left(\sqrt{\frac{27}{8}} \cdot \frac{H_e}{R} \right) \quad (8.5)$$

Stiffness K_1 of mass M_0 is given by:

$$K_1 = M_0 \cdot \omega_0^2 \quad (8.6)$$

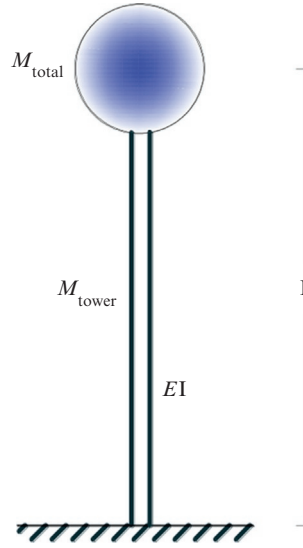
For the calculation of the spring constant K_0 , we start from the fundamental period value of a structure that can be estimated from empirical formulas, or calculated by analytical or numerical method. Empirical relations proposed by the Algerian seismic code are only applicable for buildings. In the case of an elevated tank, it can be considered as an inverse pendulum in a realistic way. For this reason, we used the Rayleigh method, which considers the elevated tank as a console. It allows the calculation of the fundamental period of the first mode of structure vibration, assimilated to a concentrated mass resting on a support of significant mass and a constant cross-section. The tank is modeled mechanically into a single mass supported by a tower (see Figure 8.4). The period is given by:

$$T = 2\pi \cdot \sqrt{\frac{P' \cdot l^3}{g \cdot 3EI}} \quad (8.7)$$

$$P' = \left(M_{total} + \frac{33}{140} \cdot M_{tower} \right) \cdot g \quad (8.8)$$

M_{total} : concentrated mass (mass of the concrete vessel and the water it contains); M_{tower} : mass of the supporting system; I : moment of inertia of the supporting system cross section; E : elasticity modulus of concrete; l : height of gravity center of the oscillating mass to the embedding.

This relation implicitly assumes that it is not coupled planar oscillations with other modes of oscillations. That is to say, there are the oscillations in which the various masses of structure move parallel to the same plane, without exciting perpendicular oscillations to this plane. This condition is satisfied by the structures having a vertical plane of symmetry, as it is in the case of circular tanks. In addition to the

**FIGURE 8.4**

Tank modeled mechanically as a single mass supported by a tower.

geometrical characteristics of the structure (dimensions of the tank), the relation uses the flexional rigidity (EI), and the weight distribution of the structure (P). These are two missing parameters in the empirical relation proposed in the Algerian seismic code.

The rigidity K_0 of the supporting system is deduced by:

$$\omega^2 = \frac{4\pi^2}{T^2} = \frac{g \cdot 3 \cdot EI}{P' \cdot l^3} = \frac{K_0}{M_{\text{total}}} \quad (8.9)$$

Then,

$$K_0 = \frac{M_{\text{total}}}{M_{\text{total}} + \frac{33}{140} \cdot M_{\text{tower}}} \cdot \frac{3 \cdot EI}{l^3} \quad (8.10)$$

The calculation of the stiffness is given in the [Table 8.6](#).

5.3.4 Calculation of eigenpulsations

The equilibrium equations can be formulated in a matrix form by the following equation:

$$\{f_i(t)\} + \{f_a(t)\} + \{f_c(t)\} = \{f(t)\} \quad (8.11)$$

The vector of inertial forces can be written as $\{f_i(t)\} = [M] \cdot \{\ddot{X}(t)\}$

Table 8.6 Evaluation of Stiffness

Fundamental pulsation of vibration w_0^2	5.95 1/s ²
The stiffness K_1	134,915.35 N/m
Elastic modulus of the concrete E	32,164,195,120.34 N/m ²
Height of the supporting system	7.15 m
Moment of inertia of the supporting system cross-section I	1.17 m ⁴
$P' = (M_{\text{total}} + \frac{33}{140} \cdot M_{\text{tower}}) \cdot g$	1,488,820.03 N
Fundamental period T	0.14 s
The stiffness K_0	302,996,650.55 N/m

The vector of damping forces can be written as $\{f_a(t)\} = [C] \cdot \{\dot{X}(t)\}$

The vector of elastic forces can be written as $\{f_e(t)\} = [K] \cdot \{X(t)\}$

And $\{f(t)\}$ denotes the vector of the external forces.

In our case, damping is assumed to be null and as we restrict ourselves only to the determination the eigenmodes of vibration, we will just build two matrices $[M]$ and $[K]$. We also assume that the movement is harmonic type, described in the form:

$$\{X(t)\} = \{a\} \cdot \sin(\omega t + \phi) \quad (8.12)$$

Since the system is free swinging (i.e., without external loading), therefore $\{f(t)\} = \{0\}$ and is undamped, therefore $\{f_a(t)\} = \{0\}$, the equation of motion can be written:

$$[M] \cdot \{\ddot{X}(t)\} + [K] \cdot \{X(t)\} = \{0\} \quad (8.13)$$

Since:

$$\{\ddot{X}(t)\} = \frac{d^2}{dt^2} \{X(t)\} = -\{a\} \cdot \omega^2 \cdot \sin(\omega t + \phi) = -\omega^2 \cdot \{X(t)\} \quad (8.14)$$

The equation is then rewritten:

$$[[K] - \omega^2 \cdot [M]] \cdot \{X(t)\} = \{0\} \quad (8.15)$$

The obvious solution is the trivial solution $\{X(t)\} = \{0\}$. This solution corresponds to the undeformed initial position of tank. As the structure oscillates, so deforms to a given time t , it will have a deformation $\{X(t)\} \neq \{0\}$.

The determinant should be equal to zero, that is to say:

$$|[K] - \omega^2 \cdot [M]| = 0 \quad (8.16)$$

The matrix $[K]$ and $[M]$ can be written:

$$[K] = \begin{bmatrix} K_{00} & K_{01} \\ K_{10} & K_{11} \end{bmatrix} \quad \text{and} \quad [M] = \begin{bmatrix} M_0 & 0 \\ 0 & M_1 \end{bmatrix}$$

Where, $K_{00} = K_1$, $K_{11} = K_0 + K_1$, and $K_{10} = K_{01} = -K_1$.

So, the calculation of the stiffness matrix elements is given in the [Table 8.7](#).

Table 8.7 Evaluation of the Stiffness Matrix Elements

$K_{00}=K_1=$	134,915.35 N/m
$K_{01}=K_{10}=-K_1=$	-134,915.35 N/m
$K_{11}=K_0+K_1=$	303,131,565.91 N/m

$$\left| \begin{bmatrix} K_{00} & K_{01} \\ K_{10} & K_{11} \end{bmatrix} - \omega^2 \cdot \begin{bmatrix} M_0 & 0 \\ 0 & M_1 \end{bmatrix} \right| = 0 \quad (8.17)$$

$$\left| \begin{bmatrix} K_{00} - \omega^2 \cdot M_0 & K_{01} \\ K_{10} & K_{11} - \omega^2 \cdot M_1 \end{bmatrix} \right| = 0 \quad (8.18)$$

The determinant will be:

$$(K_{00} - \omega^2 \cdot M_0) \cdot (K_{11} - \omega^2 \cdot M_1) - K_{01} \cdot K_{10} = 0 \quad (8.19)$$

The resolution of this equation gives us the vibration pulsations of the two main modes:

$$\omega_{1,2}^2 = \frac{1}{2} \cdot \left[\left(\frac{K_{00}}{M_0} + \frac{K_{11}}{M_1} \right) \pm \sqrt{\left(\frac{K_{00}}{M_0} + \frac{K_{11}}{M_1} \right)^2 + 4 \left(\frac{K_{01} \cdot K_{10}}{M_0 \cdot M_1} \right)} \right] \quad (8.20)$$

Knowing the pulsations of the eigenmodes ω_1 and ω_2 , we deduct the periods of the two main modes:

$$T_1 = \frac{2\pi}{\omega_1} \quad \text{and} \quad T_2 = \frac{2\pi}{\omega_2}$$

The calculation of the pulsations and periods of the two main eigenmodes is given in the [Table 8.8](#).

5.3.5 Determination of the eigenmodes

The system has a same number of degrees of freedom than eigenmodes. For the determination of the eigenmodes $\{a\}_i = \begin{Bmatrix} a_{i0} \\ a_{i1} \end{Bmatrix}$, we must solve the equation:

$$([K] - \omega_i^2 \cdot [M]) \cdot \{a\}_i = \{0\} \quad (8.21)$$

Table 8.8 Evaluation of the Pulsations and Periods of the Two Main Eigenmodes

	ω^2	Pulsation ω	Period T (s)
Mode 1	5.95	2.44	2.58
Mode 2	2,385.63	48.84	0.13

Table 8.9 Evaluation of the Two Main Eigenmodes

The mode “ <i>i</i> ”	a_{i0}	a_{i1}
Mode 1	1.00	0.0004
Mode 2	1.00	−399.99

We can also write in the following form:

$$\begin{bmatrix} K_{00} - \omega_i^2 \cdot M_0 & K_{01} \\ K_{10} & K_{11} - \omega_i^2 \cdot M_1 \end{bmatrix} \cdot \begin{Bmatrix} a_{i0} \\ a_{i1} \end{Bmatrix} = \begin{Bmatrix} 0 \\ 0 \end{Bmatrix} \quad (8.22)$$

or in this form:

$$\begin{cases} (K_{00} - \omega_i^2 \cdot M_0) \cdot a_{i0} + K_{01} \cdot a_{i1} = 0 \\ K_{10} \cdot a_{i0} + (K_{11} - \omega_i^2 \cdot M_1) \cdot a_{i1} = 0 \end{cases} \quad (8.23)$$

If we solve the system as a Cramer system, the solution that would be obtained is $a_{i0} = a_{i1} = 0$.

But, as an eigenmode corresponds to a deformed position, the solution should be different than zero. For this, we give an arbitrary value, usually equal to unity for “ a_{i0} ” and we deduce “ a_{i1} ”.

If we put $a_{i0} = 1$, we deduce $a_{i1} = -\frac{K_{00} - \omega_i^2 \cdot M_0}{K_{01}} \cdot a_{i0}$.

The results of the evaluation of the two main eigenmodes are given in the [Table 8.9](#).

5.3.6 Evaluation of the seismic force by the dynamic method of spectral modal analysis

The Algerian seismic code, allows under certain conditions, the calculation of the structure by the pseudo dynamic method which consists in considering the structure as being subjected to a shear force which is a function of several parameters. In this method, the masses are assumed concentrated at the main nodes and only the horizontal displacements of the nodes are taken into account.

The lateral seismic force applied to the mass “ k ” and mode “ i ” is given by the following relation:

$$F_{ik} = \frac{S_{ai}}{g} \cdot \gamma_i \cdot M_k \cdot a_{ik} \quad (8.24)$$

Where S_{ai} : ground acceleration at the mode “ i ”; γ_i : distribution coefficient to the mode “ i ” and is given by the following relation:

$$\gamma_i = \frac{\sum_{k=1}^n M_k \cdot a_{ik}}{\sum_{k=1}^n M_k \cdot a_{ik}^2} \quad (8.25)$$

The calculation of the distribution coefficients of the two main eigenmodes is given in the [Table 8.10](#).

Table 8.10 Evaluation of the Distribution Coefficients of the Two Main Eigenmodes

The mode “ <i>i</i> ”	γ_i
Mode 1	1.00
Mode 2	−0.00249893

5.3.7 Calculation of the acceleration S_a

The imposed acceleration “ S_a ” on the tank, considering its interaction with the ground, is a data of seismic code based on the seismic zone and other parameters. It is a fraction of the gravity acceleration. The Algerian seismic code proposes for its determination the following relations:

$$\frac{S_a}{g} = \begin{cases} 1.25A \left(1 + \frac{T}{T_1} \left(2.5\eta \frac{Q}{R} - 1 \right) \right) & 0 \leq T \leq T_1 \\ 2.5\eta(1.25A) \left(\frac{Q}{R} \right) & T_1 \leq T \leq T_2 \\ 2.5\eta(1.25A) \left(\frac{Q}{R} \right) \left(\frac{T_2}{T} \right)^{2/3} & T_2 \leq T \leq 3.0 \text{ s} \\ 2.5\eta(1.25A) \left(\frac{T_2}{3} \right)^{2/3} \left(\frac{3}{T} \right)^{5/3} \left(\frac{Q}{R} \right) & T > 3.0 \text{ s} \end{cases} \quad (8.26)$$

Where, “ A ” is the acceleration coefficient of the zone, depending on the seismic zone and the group use of the structure. The Algerian seismic code classes the storage tanks as structures of group 1B (Table 8.11).

η is a correction factor of damping given by the following relation:

$$\eta = \sqrt{7 / (2 + \xi)} \geq 0.70 \quad (8.27)$$

where ξ denotes the percentage of critical damping. Its value depends on constituent material, the type of structure, and the importance of filling (Table 8.12).

The classification of structural systems “ R ” is made taking into account their reliability and their ability to dissipate the energy to the seismic action, and the

Table 8.11 Acceleration Coefficient of the Zone “ A ”

Group	Zone I (Low Seismicity)	Zone IIa (Average Seismicity)	Zone IIb (Average Seismicity)	Zone III (High Seismicity)
1A	0.15	0.25	0.30	0.40
1B	0.12	0.20	0.25	0.30
2	0.10	0.15	0.20	0.25
3	0.07	0.10	0.14	0.18

Table 8.12 Percentage of Critical Damping ξ

Filling	Frames		Shell or Wall
	RC	Steel	RC and Masonry
Light	6	4	10

Table 8.13 Observed Penalties P_q

No	Criterion q	Factor P_q	
		Observed	Not Observed
1	Minimum requirements on lines bracing	0	0.05
2	Redundancy plan	0	0.05
3	Regularity in plan	0	0.05
4	Regularity in elevation	0	0.05
5	Quality control of materials	0	0.05
6	Quality control of the execution	0	0.10

Table 8.14 Characteristic Periods T_1 , T_2

	Type S ₁	Type S ₂	Type S ₃	Type S ₄
Site	Rocky Soil	Firm Soil	Loose Soil	Very Loose Soil
T_1 (s)	0.15	0.15	0.15	0.15
T_2 (s)	0.30	0.40	0.50	0.70

corresponding behavior coefficient is determined according to the nature of the constituent materials, the type of construction, the possibility of forces redistribution in the structure, and deformation elements capabilities in the post-elastic domain. The bracing system used in the study of this tank is the inverted pendulum system, which, according to the Algerian seismic code is a system where 50% or more of the mass is concentrated in the upper third of the structure. Algerian seismic code gives value, to global behavior coefficient of the tank, $R=2$.

The quality factor, “ Q ” is determined by the following relation: $Q = 1 + \sum P_q$.

In this relation, P_q means the penalties to remember accordingly that the quality criterion is satisfied or not (see Table 8.13). For the case of our elevated tank, $Q=1$, because the six penalties are observed.

The characteristic periods T_1 and T_2 are associated with the category of the site, given in Table 8.14.

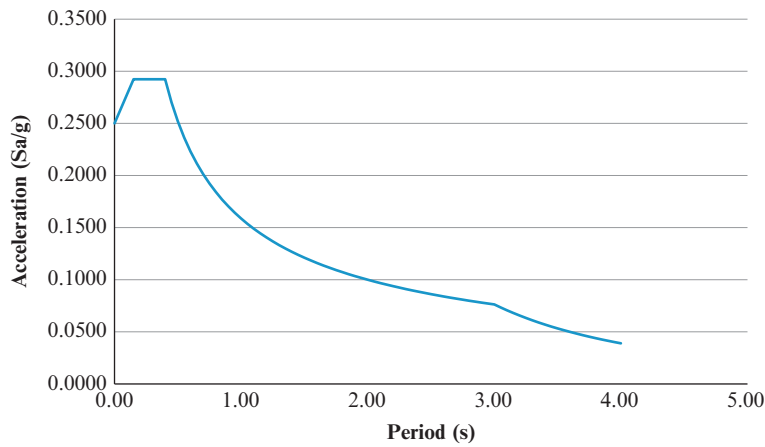
We give in the Table 8.15 all the different coefficients allowing drawing the response spectrum curve (Figure 8.5).

5.3.8 Evaluation of the safety factor

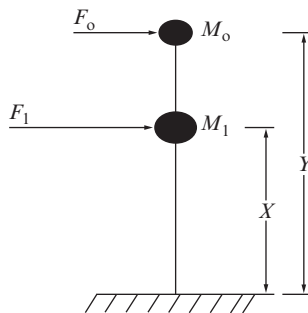
The resulting seismic force at the mass M_k is given by the mean quadratic (Figure 8.6):

Table 8.15 Different Coefficient of the Response Spectrum

Coefficient of acceleration zone $A =$	0.20
Critical damping $\xi =$	6.00
Global behavior coefficient of the structure $R =$	2.00
Characteristic period $T_1 =$	0.15 s
Characteristic period $T_2 =$	0.40 s
Quality factor $Q =$	1.00
Correction factor of damping $\eta =$	$0.94 > 0.7$

**FIGURE 8.5**

Response spectrum.

**FIGURE 8.6**

The resulting seismic force applied on each mass.

Table 8.16 Evaluation of the Resulting Seismic Force on Each Mass with Their Application Point

The mode “ <i>p</i> ”	F_{i0}	F_{i1}
Mode 1	1,932.91 kg	4.83 kg
Mode 2	–15.77 kg	35,335.31 kg
Resulting seismic force	1,932.98 kg	35,335.31 kg
Application point	12.07 m	9.98 m

Table 8.17 Evaluation of the Safety Factor

Mass of a full tank	184,380.41 kg
External radius of the foundation raft of the elevated tank	2.60 m
Stabilizing moment	4,702,806.83 Nm
Maximum overturning moment	3,688,860.33 Nm
The safety factor F_s	1.27

$$F_k = \sqrt{\sum_{i=1}^n F_{ik}^2} \quad (8.28)$$

The evaluation of the resulting seismic force on each mass (see [Figure 8.6](#)) with their application point is given in [Table 8.16](#).

Under the effect of the seismic action at ultimate limit state, the overall stability of the tank to the collapse must be satisfied. For this purpose, it is necessary to calculate the safety factor:

$$F_s = \frac{M_s}{M_r} \quad (8.29)$$

Where, M_s is the stabilizing moment.

The overturning moment is calculated by the relation:

$$M_r = F_0 \cdot Y + F_1 \cdot X \quad (8.30)$$

The evaluation of the safety factor is given in the [Table 8.17](#).

If the safety factor is less than 1.50, the tank has a poor stability against overturning under seismic action. The parameter “seismic behavior” is classified as **class C**.

5.4 SOIL TYPE

Vulnerability due to the influence of soil can have two origins, the soil instability, due to its composition, its liquefaction, and to landslides, or then the resonance effect between the soil and the height of the structure. The geological aspect is taken into account in analyzing the implementation category site which is based on mechanical properties of soil. The sites are classified into four categories by the Algerian seismic code according to the average velocity of the shear wave V_s .

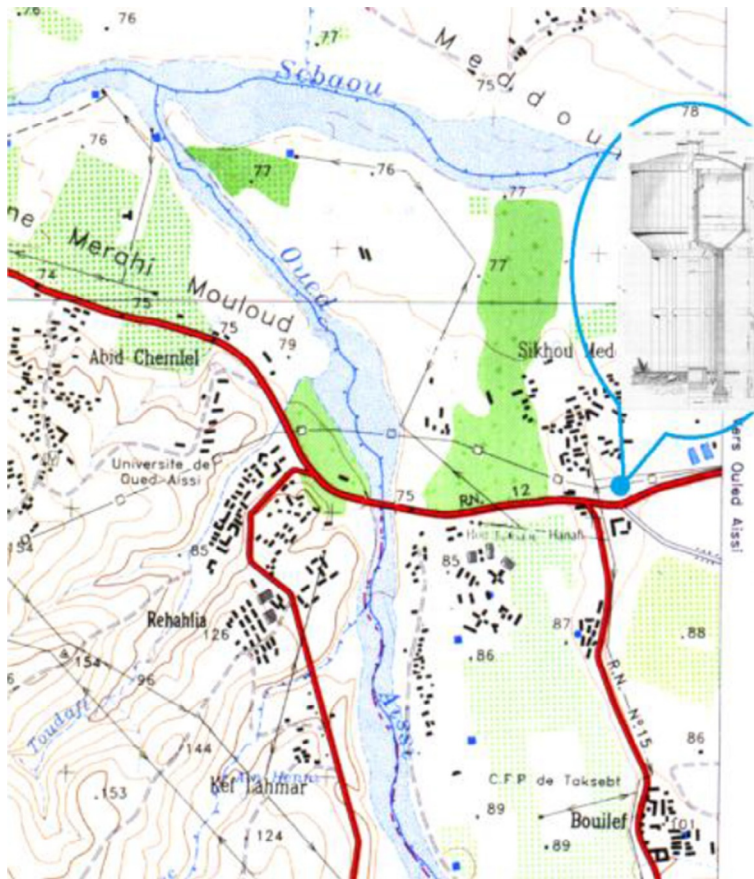


FIGURE 8.7

The general view of Sikh Oumedour tank.

The tank is located in Sikh Oumedour village, on the valley of Sebaou, bounded on the North by the main stem of the Sebaou river and on the west by a tributary named Aissi river (Figure 8.7). Reading the geological report shows that the valley is composed of quaternary alluvial deposits sitting on bedrock of tertiary Eocene marls. This alluvial layer about 50 m thick, characteristic of the middle Sebaou, consists of deposits of sand and very dense gravel. This soil is identified as soil category type S2 (firm soil), according to the Algerian seismic code. This parameter is classified as **class B**.

5.5 REGULARITY IN PLAN

The seismic movements result in overall torsion of structures when their center of mass does not coincide with the center of rigidity. Torsion is an important factor in vulnerability; its limitation should be systematically required. For this reason,

Table 8.18 Evaluation of the Projecting Dimensions Ratio

The external diameter of the vessel (D)	6.20 m
The external diameter of the supporting system (d)	4.00 m
The ratio $D - d/D$	0.35

it may be necessary to add bracing elements positioned to rebalance the distribution of rigid elements. Analysis of seismic damage, clearly shows that regular and symmetrical structures perform better than those whose geometric forms and the distribution of resistive elements are complex, even when the seismic rules were correctly applied. The calculations, as prescribed by these seismic codes, represent seismic stresses to structures of a fairly realistic way for regular structures whose seismic response is simpler than for irregular structures [15]. The symmetry and regularity in plan have the function of providing better resistance to earthquakes. Structures should preferably have, on one hand, simple forms and on the other hand, distribution as regular as possible of masses and stiffness as in plan and elevation. The aim is to ensure the best possible distribution of forces through the structure, so as to involve all the resistant elements to the absorption and dissipation of the energy developed by the seismic action. Taking inspiration of the seismic building rules [16], applied to buildings, we can consider the following criteria:

- The tank should have a symmetrical configuration in both directions, both for the distribution of rigidities than those of the masses. The inspected tank presents a perfectly symmetrical configuration in both directions, both for the distribution of rigidities than for the masses. This leads us to say that the center of mass coincides with the center of torsion (rigidities). This criterion is satisfied.
- The tank form should be compact. The dimensions of the projecting parts shall not exceed 25% of the total tank size in the corresponding direction in accordance with this relation: $\frac{D-d}{D} < 0.25$ (see Figure 8.8). In our case, the ratio of the projecting dimensions is greater than 25% (see Table 8.18). This criterion is not satisfied (Figure 8.8).
- The geometric slenderness ratio λ_g of the supporting system, which defines the ratio between the buckling length and the minimum gyration radius of the cross-section of the support, should be between the following values: $3 \leq \lambda_g \leq 5$. In our case, this criterion is satisfied (see Table 8.19).

In conclusion, the regularity in plan is partially verified. This parameter is classified as **class B**.

5.6 REGULARITY IN ELEVATION

The principle of regularity in elevation is to avoid discontinuities or excessively fast changes in the distribution of inertias, stiffness, and masses. Taking inspiration from the seismic building rules [16], applied to buildings, we can consider the following criteria:

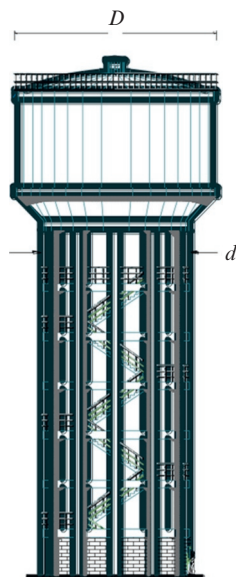


FIGURE 8.8
Different forms of elevated tanks.

Table 8.19 Evaluation of the Geometric Slenderness Ratio λ_g of the Support

Height of the supporting system	7.15 m
The buckling length	5.01 m
Minimum moment of inertia of the supporting system cross-section	1.17 m ⁴
Cross-section of the columns supporting system	0.72 m ²
The gyration radius of the cross-section of the columns supporting system	1.28 m
The geometric slenderness	3.92

- Mass distribution between the vessel and the support must be substantially regular over the height of the structure, the mass ratio being between the following values: $0.63 < \frac{M_c}{M_{tower}} < 1.33$. Where, M_c means the mass of the container (vessel) and M_{tower} means the mass of the supporting system. In our case, this criterion is not satisfied (see Table 8.20).
- The stiffness distribution should be substantially regular over the height of the structure. In our case, the piles of the supporting system have a constant inertia all along the height. This criterion is satisfied.

In conclusion, the regularity in elevation is partially verified. This parameter is classified as **class B**.

Table 8.20 Evaluation of the Mass Ratio

Mass of the empty vessel M_c	44,327.83 kg
Mass of the supporting system M_{tower}	12,870.00 kg
The ratio M_c/M_{tower}	3.44

5.7 QUALITY OF THE NODES

We define the nodes by the connection areas between the structural elements, such as the connecting zone “Foundation—support” and the connecting zone “support—tank.” Among several reasons that are at the origin of failures and damage caused by earthquakes, we cite poor design and construction of the nodes in tanks between resistant elements. This observation has been demonstrated in previous studies by Masoudi et al. [17]. The state of connecting zone “Foundation—support” and the connecting zone “support—tank” is good, without cracks. This parameter is classified as **class A**.

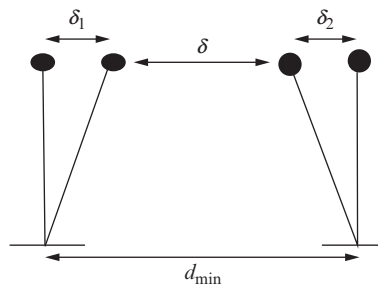
5.8 SEISMIC ZONE

The tank is located in Sikh Oumedour village, municipality of Tizi Ouzou. The Algerian seismic code classes this municipality as average seismicity zone (zone IIa) (see Table 8.11). This parameter is classified as **class B**.

5.9 THE CLASHING

The clashing is the interaction between two or more adjacent tanks whose effect of the seismic behavior, is one relative to the other, during the seismic occurrence is important. The solution to this problem is to create a mechanical joint. The classification of this parameter is a function of the minimum width of seismic joints, given by the following relation: $d_{\min} = \delta + \delta_1 + \delta_2$.

δ_1 and δ_2 , are the maximum displacements of the two adjacent tanks (Figure 8.9) and δ is the residual distance between the two adjacent tanks. In our case, the study

**FIGURE 8.9**

The minimum width of the joint.

tank is isolated. So, there is no risk of clashing. This parameter is classified as **class A**.

5.10 SITE EFFECT

Tanks may be threatened by some effects of the implantation site (as liquefaction of sands, landslides, presence of surface faults, and rockfalls) that can significantly amplify the oscillations of the soil and cause the loss of all structure even para-seismic. Structures located there can sometimes undergo seismic loads up to five times higher than similar structures located in a less dangerous zone [4]. It is therefore imperative to take into account the site effect in a vulnerability study and the resort to a geotechnical engineer or geologist is desirable. Our tank is located in site that presents no risk. So, this parameter is classified as **class A**.

5.11 DETAILS

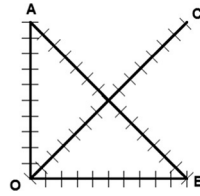
We often hear about the detail parameter as the state and quality of nonstructural elements that can influence the tank behavior during an earthquake and the status of various networks related to the functionality of the tank. The elements to consider for this parameter are the state of hydraulic equipment (various distribution pipelines, etc.), the cover roofing, the cornice, and the stairs. All nonstructural elements are in bad condition. This parameter is classified as **class C**.

5.12 MAINTENANCE

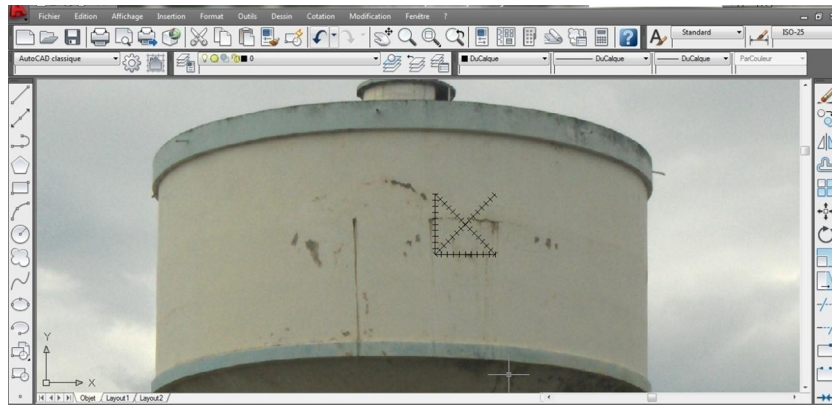
Cracks on resistant components of the tank shall result in the loss of stability and loss of the sealing function. Quantifying the cracking state is made according to the LPC method [18] which is a quick and easy way to characterize in conventional manner of the state of reinforced concrete wall damage at a given time. This method concerns all existing cracks in a given area. The method consists of taking in an exhaustive manner all cracks intercepting four axes drawn in an area of one square meter on the part of the facing to study. A landmark consisting of four axes graduated from 0.1 to 0.1 m (a vertical axis of 1 m length, a horizontal axis of 1 m length, and two axes arranged at 45° relative to the horizontal axis with 1.4 m length) is drawn on the facing using the jig (Figure 8.10). The tracing is performed carefully to avoid altering the lips of the cracks.

Cracking zone is shown on the cylindrical wall, where seepage is observed, on which we have drawn the landmark with four axes (see Figure 8.11).

All the cracks with openings greater than 0.05 mm, intersecting with the axis of the landmark are identified and measured up to 2 mm opening. For each of the four axes, we calculate the total accumulated of the opening, the average opening per crack, and the average opening per meter length axis. The cracking index is calculated by taking the average of the four openings per meter obtained on each axis. The detail of this calculation is given in Table 8.21.

**FIGURE 8.10**

The four axes of the landmark.

**FIGURE 8.11**

Drawing of the landmark on the cracking area.

According to the [Table 8.22](#), the importance of cracking is considered moderate. And, as the cracking index is between 1 and 5, this parameter “maintenance” is classified as **class B**.

5.13 DETERMINATION OF THE VULNERABILITY INDEX

The vulnerability index value of the tank calculated according to the [Table 8.23](#), classes the tank vulnerability in the Orange level 1. The tank has an average vulnerability to seismic risk. A repair program is necessary to increase its seismic strength.

Table 8.21 Calculation of the Crack Index

Interval	1	2	3	4	5	Basic Length (m)	Number of Cracks	Cumulated Total Opening	Average Opening by Crack	Average Opening by Meter	Overall Average Opening
	6	7	8	9	10						
	11	12	13	14							
OA		2				1.00	1.00	2.00	2.00	2.00	1.71
OB	2					1.00	1.00	2.00	2.00	2.00	
OC				2		1.40	1.00	2.00	2.00	1.43	
AB				2		1.40	1.00	2.00	2.00	1.43	

Table 8.22 Characterization Scale of Cracking

Value of Crack Index	Importance of Cracking
0-0.5	Negligible
0.5-1	Low
1-2	Moderate
2-5	Strong
5-10	Very strong
>10	Considerable

Table 8.23 The Vulnerability Index " I_V " Calculated for the Tank Investigated

No	Parameter	Class	K_i	W_i	$K_i W_i$
1	Seismic behavior	C	15.00	1.75	26.25
2	The clashing	A	—	1.75	-
3	The bracing system	C	15.00	1.50	22.50
4	Regularity in plan	B	5.00	1.50	7.50
5	Regularity in elevation	B	5.00	1.50	7.50
6	Soil type	B	5.00	1.00	5.00
7	Site effect	A	—	1.00	-
8	Seismic zone	B	5.00	1.00	5.00
9	Quality of the nodes	A	—	0.75	-
10	Qualities of the bracing system	B	5.00	0.75	3.75
11	Maintenance	B	5.00	0.25	1.25
12	Details	C	15.00	0.25	3.75
				I_V	82.50

6 CONCLUSION

The method developed is used to establish a hierarchy of tanks with risks from pre-defined safety criteria. The decision on the choice of maintenance, accrued to the manager of the tanks, consists in fact to estimate the limit vulnerability, which leads them to choose preventive maintenance (short-term repairs in order to prevent further deterioration to come) or consider a deferral of intervention if the tank does not represent high risk in terms of safety or cost, or without repairs.

According to the principles used for the vulnerability index evolution in the life-cycle of the tank for the establishment of maintenance schemes, two terminals of vulnerability will be defined (I_{V1} and I_{V2}) segmenting all of the scale in three categories maintenance:

- Tanks whose vulnerability index is less than 33 (I_{V1}): no maintenance is planned; only a regular monitoring (visual inspection, auscultation) is considered.

- Tanks whose vulnerability index is greater than 133 (I_{V2}): the state of the tank is considered sufficiently degraded to justify short-term repair.
- Tanks whose vulnerability is between I_{V1} and I_{V2} : additional studies and analysis of the potential risks of worsening defects must be engaged to decide on the nature and urgency of the repair work to be undertaken.

In the hands of managers, this method allows them to fix the priorities for action in their rehabilitation program or repair, in order to optimize its management under significant budgetary constraints.

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Reliability and material failure analysis of water and wastewater systems: Case studies

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CHAPTER OUTLINE

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1 WATER SUPPLY AND WASTEWATER DISPOSAL INFRASTRUCTURE

In Poland, over 283,000 km of water supply networks and about 50,000 km of service lines were in operation in 2012. In the same year, wastewater disposal networks amounted to 126,000 km [1]. A dynamic development of networks has been observed in Poland since 1990: the average of about 4500 km of water supply pipelines and over 8000 km of wastewater disposal pipelines were constructed annually in the period 2005-12. In the Czech Republic, 21% (2012) of the population was supplied with water from public water supply networks, while 79% receives water from non-public water supply networks. In Canada, the length of water supply networks reaches almost 91.1% [2].

Materials used to manufacture water supply or wastewater disposal conduits should be characterized by the following properties:

- Leak-tight
- Inert relative to the transmitted medium (water or wastewater)

- Resistance to corrosion and aggressive compound action
- Durable
- Resistant to temperature changes, both of the external environment and of the transferred medium
- Resistant to shape deformation
- Resistant to abrasion
- Ease of manufacture and assembly
- Homogenous molecular structure
- Inexpensive to manufacture, implement, and operate
- Readily available on the market

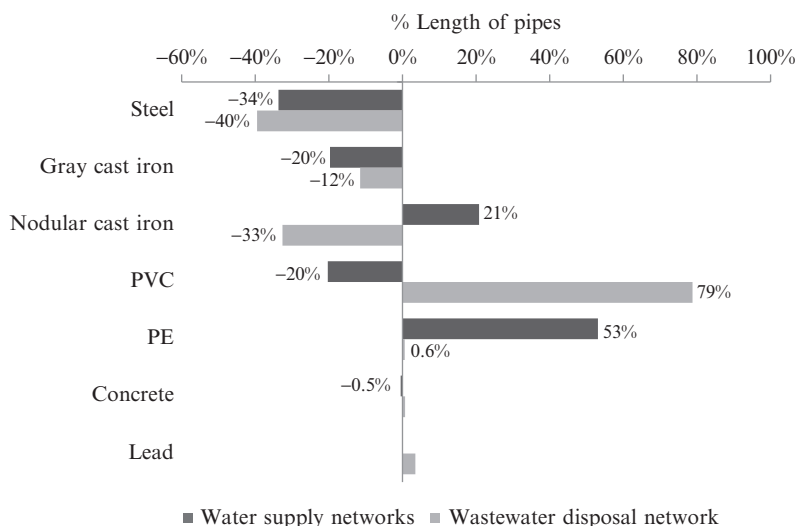
The above requirements concerning the quality of materials to be used may be extended to cover additional features that are related to specific local conditions. In the context of a broad assortment of products and materials considered, particular significance should be ascribed to the problem of how to select a material that is best fitted to a planned investment in network infrastructure. Water supply and wastewater disposal infrastructure is found in every public location, urban agglomeration, rural housing area, building, facility, industrial sector, or branch of industry. At the same time, water supply systems and wastewater disposal systems play a very important role in terms of reliability of water supply and wastewater disposal. As they belong to complex network systems, they simultaneously fulfill several functions within a given area. Large-scale systems have been functioning in an organized manner for over 140 years in various countries. Vulnerability of such systems to failure has been investigated in Poland for over 50 years [3–9].

Investigations cover analyses of types and causes of damages, as well as identification of various reliability parameters. The range of investigated objects is highly differentiated and covers, for instance, conduits of municipal and rural water supply networks; transit, arterial, distribution conduits; household connection lines; conduits operating in various soil conditions (e.g., in zones affected by mining damages or increased risk of seismic activity); conduits made of a variety of materials. In the case of wastewater disposal networks, the investigations cover collector pipes, network conduits, house drains, transfer conduits or small-sized, local wastewater disposal systems. The research also covered network systems located in a variety of industrial facilities, both in small-sized plants and in large-scale factories.

About 50% of the length of the water supply networks in Poland is still made of traditional materials, yet water supply and wastewater disposal conduits are no longer made of steel, while gray cast iron is replaced mainly by nodular cast iron or plastics.

Figure 9.1 presents changes that occurred between 1992 and 2005 in the distribution of percentage shares of particular materials in the examined water supply networks.

It should be underscored that small diameter conduits (primarily DN 100 and DN 150) constitute a clear majority of water supply networks. In big cities, conduits that are 100 and 150 in diameter cover about 50% of the length of municipal networks.

**FIGURE 9.1**

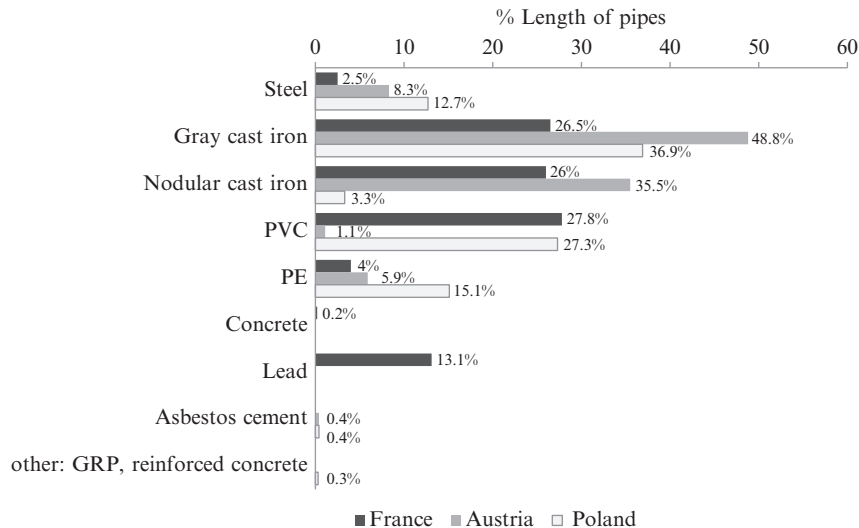
Percentage share of the length of water supply and wastewater disposal networks in 2005, compared to 1992.

In towns, the share of such conduits is even higher, while in small towns, the diameters of water supply conduits generally do not exceed the diameter of distribution networks.

Conduits made of traditional materials—such as gray cast iron—still hold a majority share in the construction materials of water supply networks both in Poland and in other countries. Conduits made of nodular cast iron constitute a considerable part of water supply networks in Austria (about 36% of the network length) and in the French region (about 26%), while in Poland, they constitute on average about 3.3% of the length of water supply networks, which is due to the fact that this material started to be implemented only in the 1990s. A high share of conduits made of plastic is found in Poland and France, while in Austria, they constitute barely 7% of the length of the network (Figure 9.2). Occasionally, one can still find conduits made of such materials as lead (already excluded from water supply conduit production) or asbestos cement that are effectively replaced with materials most often presently implemented by polyethylene (PE) or polyvinyl chloride (PVC).

2 MATERIAL AND AGE STRUCTURE OF UTILITY NETWORKS

The development of the water supply and wastewater disposal networks has been accompanied for almost 30 years by a systematic technological progress in the domain of manufacturing of products made of various materials and designed for these networks. Both modifications of traditional materials and completely new generations

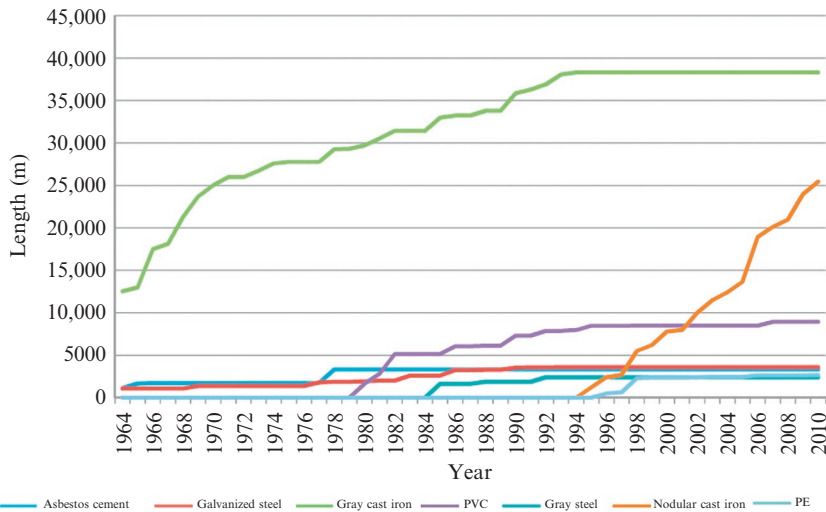
**FIGURE 9.2**

Materials of construction for selected water supply networks between 2005 and 2008 in France, Austria, and Poland.

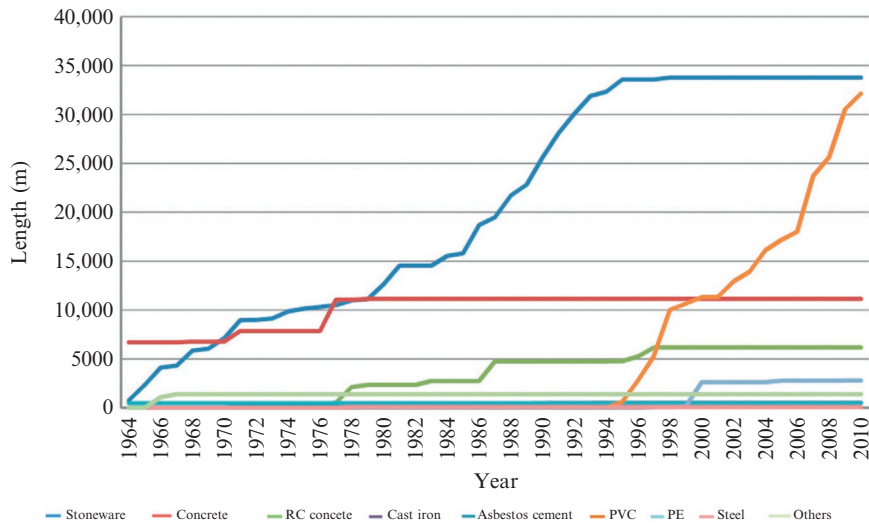
of products are available. In the context of such a broad assortment of available products and materials, particular significance should be ascribed to the problem of how to select a material solution that is best fitted to a planned investment.

It can be seen that various materials in various proportions are used to construct water supply networks. Practically since 1990, the material structure of both water supply and wastewater disposal systems has been changing. One of the biggest water supply systems in Poland is made predominantly of conduits made of gray cast iron, nodular cast iron, and steel. Thermoplastics are implemented to a very low extent. A higher share of plastic conduits can be seen in other water supply networks. There are also municipal networks, where entire systems are made exclusively of a single material, for example, PVC or PE. In many analyses, one can notice a repeatable tendency: the bigger a city, the higher the share of installations made of gray cast iron, while the smaller a water supply network, the higher the share of conduits made of PVC. A similar tendency is noticeable for wastewater disposal systems [10,11].

Water supply networks that are in operation at present are made mainly of such materials as gray cast iron and steel. Such networks constitute almost 50% of the total length of water supply networks, yet their technical condition is not good. Since 1978, apart from traditional materials, more modern solutions—such as PVC and PE (they were implemented in over 42% of the newly constructed networks in 2010)—have entered into the material structure of the water supply network, which consequently reduces the share of traditional materials. Nodular cast iron takes a smaller share in the material structure [12,13]. Therefore, one witnesses a drop in the shares taken by conduits made of traditional materials, such as steel and gray cast iron (Figures 9.3 and 9.4).

**FIGURE 9.3**

An example of cumulative construction materials of the water supply network in the period 1964-2010 [14].

**FIGURE 9.4**

An example of cumulative construction materials of the wastewater disposal network in the period 1964-2010 [13].

In the case of wastewater disposal networks, it can be seen that stoneware pipelines appear to be the dominant solution (about 50% of the network length) in the period 1992-95. However, PVC conduits have been implemented more and more often since 1996. At present, almost 90% of newly constructed wastewater disposal networks are assembled of such conduits.

The present state of the construction materials results from a considerable heterogeneity of solutions implemented in cities, towns, villages, and different regions. There are places, where water supply or wastewater disposal networks are made of only one or two types of materials. On the other hand, in some towns, especially bigger cities, these networks were and still are made of practically all available materials. It can be concluded that the use of these construction materials result from different preferences of different water supply and wastewater disposal enterprises at the moment of selecting material solutions (Figure 9.5). The preferences are often-times related to local conditions and habits of network operators that most probably result from their experience gathered up to now in the domain of system operation. They are less related to analyses of reliability or failure frequency, based on analyses of actual incidents.

The presented materials used in water supply and wastewater disposal networks reflect world-wide tendencies in the technology of production or application of these

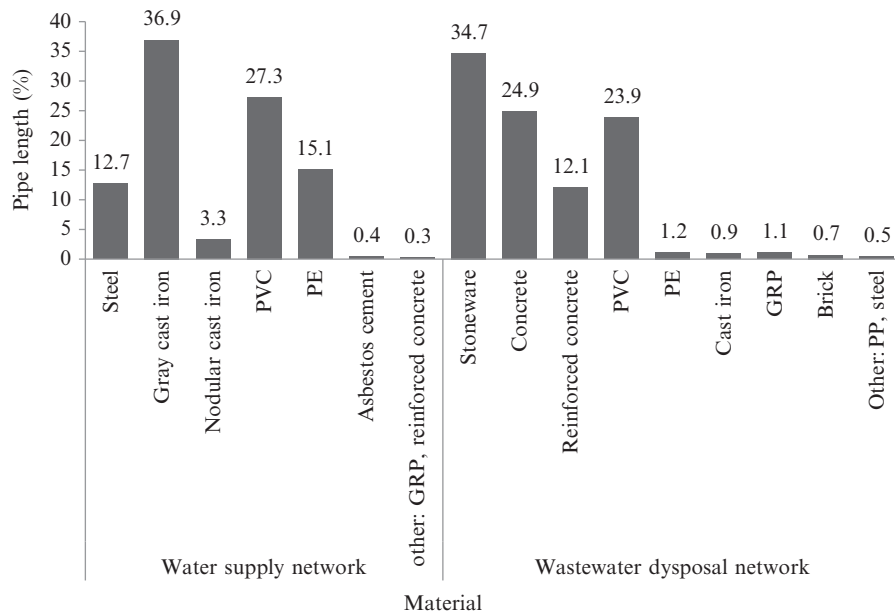


FIGURE 9.5

Materials utilized in water supply and wastewater disposal networks in 2008.

material and structural solutions to construct water supply and wastewater disposal pipelines. As construction of particular systems proceeded over many years, it is hard to determine unequivocally the age structure of the networks. On the basis of general data and historical information, one can state that water supply conduits that have been functioning for over 50 years in many big cities are primarily made of gray cast iron or steel, and less frequently of asbestos cement. In wastewater disposal systems, the highest percentage of conduits that are over 50 years old are networks made of stoneware and concrete, while a share of reinforced concrete appears in conduits that are 25-50 years old. On the other hand, in the majority of these cities, conduits that are 25-50 years old belong to the first generations of conduits made primarily of PVC and constitute 30-45% of the network length. Relatively, new conduits in wastewater disposal systems, that are 15-25 years old, are conduits made of PVC and PE. Material analysis combined with age structure analysis allows one to conclude that the service durability of materials used to construct water supply and wastewater disposal networks suggests a considerable part of the conduits is in need of replacement (Table 9.1).

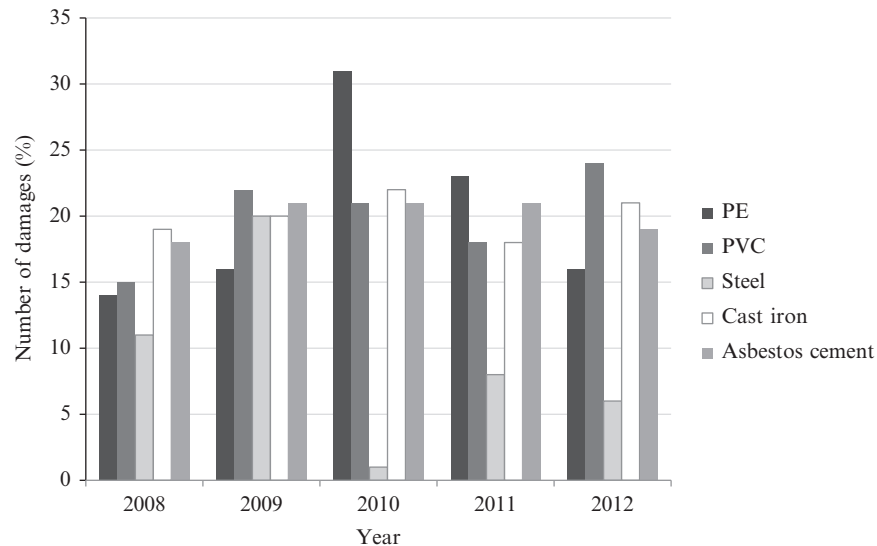
The Conduit age is an important factor that influences system susceptibility to failure, as it is correlated with conduit service durability. More vivid data illustrating the dependence of the age and material type refer mainly to all the wastewater disposal systems that were assembled as early as the nineteenth and twentieth centuries. However, not all materials show this correlation.

In order to investigate the range of changes in the number of damages as a function of time, the analysis focused on variability of the number of damages suffered in successive years of observation. The investigation took the type of conduit materials into account. Results are presented in the diagram in Figure 9.6 and Table 9.2.

In the investigation period, the number of failure events tends to grow. Years 2010 and 2011 are exceptional, as considerable differences in the numbers of failures were identified in these years. The highest number of damages (415) was identified in 2010, while only 351 network operation disturbances were recorded the next year. The above situation was primarily due to failures of cast iron conduits. One can also notice that the number of failures tends to grow in the case of all conduits, irrespective of their material. However, the highest increase is observed in the case of

Table 9.1 Material Type and Suggested Service Durability as a Function of System Type

Age (Years)	Water Supply	Wastewater Disposal Network
60-100 [15,16]	Steel	Brick
60-120 [15,16]	Gray cast iron	
≥50 lat [17]	Gray cast iron, steel, asbestos cement	Stoneware, concrete, brick
50-25 [17]	PVC	Reinforced concrete
25-15 [17]	PE, nodular cast iron, PVC	PVC, PE

**FIGURE 9.6**

The distribution of damages of water supply conduits made of different materials in the period 2008-12, broken down into particular materials.

thermoplastic materials. The analysis covered pipelines made of cast iron, asbestos cement, steel, PVC, and PE. Failures of pipelines made of other materials were disregarded, due to lack of sufficient data.

3 DAMAGE ANALYSIS

Susceptibility to failure during service life (hereinafter referred to as susceptibility to failure) is one of the crucial features that should be taken into account, when material and structural solutions are selected for the construction of water supply and wastewater disposal conduits. This feature is simultaneously the measure of conduit reliability. Thus, susceptibility to failure represents conduit failure frequency that is better recognized and more frequently used by practitioners dealing with utility network systems. According to the theory of reliability of water supply and wastewater disposal systems, system reliability is assessed by means of the so-called failure-free operation that is expressed, among others, by the failure rate. Both parameters are mutually equivalent and assume constant values in a given period of time if it is assumed that the damage flow follows Poisson distribution. This assumption was verified through service investigations focused on actual networks during their operation in regular working conditions. That is why specialists commonly use the notion of susceptibility to failure or conduit vulnerability expressed with the coefficient of failure rate. The lower the failure rate (i.e., the lower the susceptibility to failure), the

Table 9.2 Distribution of the Number of Damages of the Water Supply Network (Broken Down into Conduit Categories According to Their Function) During the Investigation Period

Year	Number of Damages											
	Type of Water Supply			Material								
	A	D	Σ	PE	PP	PVC	S	RC	CI	AC	W	Σ
2008	13	350	363	9	2	41	11	1	271	24	0	359
2009	10	380	390	10	2	57	20	2	275	23	1	390
2010	15	399	415	20	0	56	1	1	303	34	0	415
2011	15	336	351	15	2	47	8	0	251	28	0	351
2012	15	389	404	10	2	64	6	0	297	25	0	404
Σ	68	1854	1923 ^a	64	8	265	46	4	1397	134	1	1919 ^a

A, arterial conduits; D, distribution conduits; PE, polyethylene; PP, polypropylene; PVC, polyvinyl chloride; S, steel; RC, reinforced concrete; CI, cast iron; AC, asbestos cement; W, winidur (trade name of hard PVC).

^aThe above-presented values do not sum up, as no complete data concerning failure events are available.

higher the reliability. This index is credible, because it can be assessed on the basis of data collected during system service.

The data required to carry out the analysis can be broken down into two groups: data concerning damages and data identifying damaged objects. The first group covers the date and time of a failure event, duration of the failure, type of failure, and its result. The second group covers data concerning the damaged element: the conduit type, its location, the material, or diameter of a damaged conduit.

The type, effect, and cause of the susceptibility of a conduit to failure result not only from the method of its utilization, but they often result from errors in workmanship as well as in design. Apart from that, damages may be caused by internal causes, that is, originating inside a conduit, or external ones that operate from outside of the network. However, the effects of all such causes that take the final form of conduit damages generally become manifest at the final stage, that is, during the process of conduit operation. Assessment of susceptibility of conduits to failure is based on operational tests of their reliability that take the form of a natural experiment, carried out in real conditions of networks functioning and operational. However, analyses of data concerning susceptibility to failure often face the problem of absence or incompleteness of relevant data. This is related to various deficiencies of datasets, such as impossibility of distinguishing between conduits made of similar materials, missing updated data on failure duration, uncertain identification of a failure event yielding an attempted interpretation. Different situations are described utilizing the same code, which makes it impossible for a failure event to be properly classified. Moreover, it is impossible to determine the age of damaged conduits, etc. [14,18,19].

As the data concerning damages is incomplete, there arises a need to create separate subsets of data required for particular analyses. Therefore, particular analyses are based on datasets that consist of different numbers of elements and do not have to balance the datasets of higher order.

Generally, it is proposed to classify the factors that influence susceptibility of conduits to failure into the following categories according to [14]:

A: conduit-related factors:

- Assembly quality (especially the quality of connections between conduit sections)
- Material and method of anticorrosion protection
- Function and diameter
- Number and type of connections and fittings
- Age
- Pressure and pressure fluctuations (water supply system); fill-level (wastewater system)
- Water quality (corrosively); corrosiveness, aggressiveness (wastewater)

B: factors related to conduit environment:

- Soil type, aggressiveness, and humidity
- External loads (dynamic and static)
- Ground instability (seismic or mining damage)

- Temperature fluctuations
- Age

C: factors related to conduit operation:

- Maintenance, overhaul, and repair activities
- Operational procedures
- Network monitoring
- Identification and location of failures
- Failure removal

Anticorrosion protection that appeared among conduit protection factors refer primarily to conduits and elements made of metal or concrete. In the case of water supply conduits made of thermoplastics, one can observe the phenomenon of bacteria proliferation on conduit walls. In the case of metal conduits, one observes surface corrosion affecting considerable areas inside and outside a conduit, as well as local corrosion that assumes various forms, such as corrosion pits, crevice corrosion, intergranular corrosion, etc. Both in water supply conduits and in wastewater disposal conduits, one observes the phenomenon of microbially induced corrosion, which leads, among others, to the formation of deposits inside conduits. The phenomenon of corrosion in water supply and wastewater disposal conduits is correlated with numerous factors that have an impact on conduit reliability. In the water supply network, these factors include the quality of water transferred through a conduit, the type and humidity of the soil and the quality of maintenance activities performed. In a wastewater disposal network, they are related to phase interfaces between conduit material, air, and the flowing liquid, to the aggressiveness of sewage, to the amount, quality and type of gases generated in sewers as well as to the performance of pipeline operation activities. Corrosion sources may already appear at the stage of conduit assembly if pipes were damaged during transport, if conduit assembly works were not performed properly or if anticorrosion insulation had been damaged or if connections are not made correctly.

When analyzing susceptibility of a conduit to failure in terms of its impact on water quality, one should consider the corrosive impact of water on conduit material and the impact of the material on water, as these processes are interconnected. As a result, a corroded conduit is susceptible to damages, while such situation also enhances secondary water contamination that manifests itself through the deterioration of organoleptic, physical, chemical, and microbiological properties of water transferred through a conduit.

The standards prepared for the water supply and wastewater disposal sector [20–22] assume that 25 years is the technical service life of conduits in a water supply network (transit, arterial, and distribution conduits).

According to these standards, the criteria for selecting conduit service life should take numerous factors into account, including:

- Application: aggressive sewage versus fresh water
- Materials of manufacture: resistance to corrosion
- Local conditions: soil humidity, soil and water conditions
- Monitoring of operating and technical conditions

Theoretically, as a result of material wear and constant impact of physical and chemical factors as well as internal and external loads, mechanical properties of a conduit can be expected to undergo natural deterioration in time, thereby increasing conduit susceptibility to failure. This rule, however, is not confirmed by observations. An increase of the surface roughness of internal walls is also an important factor related to the actual service life, although this correlation is not unequivocal. Much research attempted to demonstrate a correlation between susceptibility of conduits to failure and their actual age, yet the results were not apparent.

It is worth pointing out that it is difficult to unequivocally determine the impact of all the factors on the susceptibility of conduits to failure, which is due to the fact that these factors most often operate jointly and generate interdependent chain of events. It is helpful to use GIS databases that allow one to create datasets that describe the status of damages, their causes and effects, as well as to collect indispensable documentation that allows one to characterize failures in a more detailed manner.

The analysis of susceptibility to failure attempts to consider the most important factors that influence the susceptibility of the water supply and wastewater disposal networks to failure. The results of research acquired up to now provide an overview of the susceptibility of conduits to failure, correlated with the type of implemented materials, as the acquired failure rates reflect a cumulative impact of various factors on the susceptibility of a single element (e.g., a conduit) to failure.

4 RELIABILITY OF WATER SUPPLY AND WASTEWATER DISPOSAL INFRASTRUCTURE

Conduits of both water supply and wastewater disposal networks operate in highly variable hydraulic conditions. This fact contributes to an increase in their susceptibility to failure, as well as to the diversity of conduit damages to be faced, compared to transit (transmission) conduits.

In the case of steel pipes, the following defects were often identified: cracks, perforations, and connection leakages. Similar defects were identified in conduits made of gray cast iron: connection leakages were identified more often in higher diameter conduits, while more cracks and perforations were identified on smaller diameter conduits. PVC conduits—both pipes and fittings—suffered most often from longitudinal and transversal cracking. Failures of hub and spigot joints were found to suffer damages at equal frequency (due to defective seals). In the case of PE conduits assembled at the beginning of the 1990s, the research identified weld cracks and sporadic fitting cracks. [Table 9.3](#) presents the most frequent failures occurring in a water supply network, as correlated with implemented materials.

Susceptibility of conduits to failure is clearly related to the implemented structural and material solution that is defined by the used material and the implemented conduit connection method.

In [Table 9.4](#), one can also notice a widespread of the failure rates of water supply conduits made of the same material and installed in different towns. This is due to

Table 9.3 Typical Failures of Conduits of Water Supply Networks Made of Different Materials [14]

Type of Failure	Failures Occurring in Conduits Made of a Particular Material			
	ST	GCI	PVC	PE
Connection leaks				
hub and spigot joints		+++	+	
welded joints	+			+
Transversal, longitudinal cracks	+	++	+++	+
Wall perforation	++	+		

+++ very frequent; ++ frequent; + rare; ST, Steel; GCI, Gray cast iron.

Table 9.4 Average Values of the Failure Rates of Water Supply Conduits Made of Different Materials and Installed in Selected Towns/Countries [17,23]

Town/Country	Average Values of the Failure Rates				Notes
	Steel	Gray Cast Iron	PVC	PE	
Czestochowa [24]	0.35	0.35	0.12	0.69	Data from 1981 to 1990 diameter ≥ 300 mm
Warsaw [25]	1.05	0.61	0.25	-	Data from 1992 to 1996 diameter ≤ 300 mm
Klodzko [26]	0.21	0.22	0.27	-	Data from 1997 to 1999; $80 \leq \text{diameter} \leq 400$ mm
Budapest [27]	0.94	0.27	0.67	-	Water supply network with household connections
Erfurt area (Germany) [26]	1.88	0.47	0.17	0.08	Data from 1980 to 1997
Hamburg (Germany) [18]	0.13	0.06	0.006	-	
Sweden [28]	0.47	0.19	0.1	0.018	
Average	0.72	0.31	0.23	0.26	

differences in the conditions of assembly and operation of investigated networks, which corresponds to a wide range of factors that influence conduit susceptibility to failure. The table below presents a comparative schedule of average values of the failure rates of water supply conduits made of different materials and installed in selected towns.

The highest susceptibility to failure was identified in the case of steel conduits that are usually the oldest ones, while the lowest failure rate was identified in the case of conduits made of thermoplastics. The research carried out in Poland [10] yielded average values of failure rate within the range of 0-0.69 failures/km.

Operational reliability investigations showed a correlation between failure rate and conduit diameter. One can notice that failure rate drops as the conduit diameter increases, which is usually correlated with the conduit function. This fact allowed researchers to formulate a generalized formula describing the correlation between failure rate and conduit diameter, irrespective of the implemented material (Figures 9.7 and 9.8) [9]:

$$\lambda_p = 99.94259 \cdot d^{(-0.99534)}$$

where: λ_p —failure rate (failure/km·year); d —conduit diameter (mm).

It is difficult, however, to determine objectively a satisfactory and acceptable level of susceptibility of conduits to failure. It turns out to be important to accept standards that determine the criteria of a unit failure rate that would provide a basis for assessing reliability resulting from an analysis of susceptibility to failure.

To assess particular material solutions in terms of susceptibility to failure and reliability, a scale from 1 (high susceptibility to failure/low reliability) to 10 (low susceptibility to failure/high reliability) can be used. In order to facilitate an assessment of material solutions by means of the scale, it is proposed to adopt criteria-delimiting values of failure rate (λ) that group conduits into three categories of susceptibility to failure/reliability (Table 9.5) and generalized failure rate values resulting from research on susceptibility to failure (Table 9.6) [17].

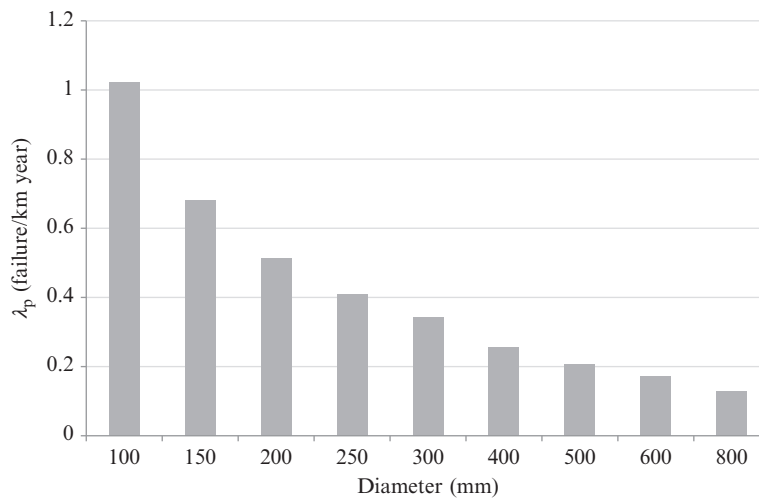
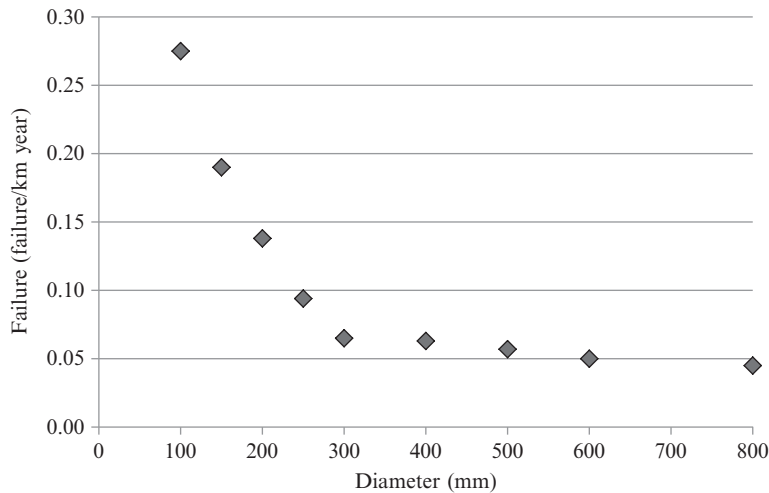


FIGURE 9.7

Correlation between failure rate and conduit diameter according to the formula.

This correlation is confirmed by the results of British research [15] quoted by [29].

**FIGURE 9.8**

Susceptibility of nodular cast iron to failure as a function of conduit diameter.

According to British Bodycote PDL data [15] quoted by [29].

Table 9.5 Classification of Water Supply Conduits in Terms of Susceptibility to Failure/Reliability (According to Kwietniewski and Rak [14])

No	Failure Category	Reliability Category	Failure Rate (failure/km-year)
A	Low susceptibility to failure	High susceptibility to reliability	$\lambda_A \leq 0.1$
B	Average susceptibility to failure	Average susceptibility to reliability	$0.1 < \lambda_B \leq 0.5$
C	High susceptibility to failure	Low susceptibility to reliability	$\lambda_C \geq 0.5$

Precise point values of conduit assessment in terms of its susceptibility to failure within the proposed scale should be assumed, when the conditions of existence and operation of assembled conduits are known. This includes known impact of various factors on susceptibility of conduits to failure. An assessment of susceptibility to failure will depend highly on the long-term experience of researchers and other entities that have been operators of conduits made of a particular material. A high value within the scale can be ascribed if conduits are characterized by the coefficient of susceptibility to failure that is not higher than 0.1 failure/km-year. Otherwise, particular material solutions are proposed to be ascribed lower values within the scale.

Table 9.6 Generalized Values of the Failure Rates of Water Supply Conduits as a Function of Conduit Material, Based on Data Collected in 2008 (According to Kwietniewski and Rak [14])

Material	Failure Rate (failure/km-year)
	Water Supply
Gray cast iron	0.47
Nodular cast iron	0.08
Steel	0.83
PVC	0.09
PE	0.08
Concrete	0.29

5 EXAMPLES OF SERIOUS FAILURES OF WATER SUPPLY AND WASTEWATER DISPOSAL SYSTEMS: CASE STUDY

The factors most often leading to failure of utility network systems include: lack of employment of protection methods, lack of protective devices, lack of preventive measures, or faulty operation resulting from operational errors [6,19,30]. Serious failures occur frequently; for instance, the last one that occurred on 30th July 2014 at the *University of California, Los Angeles*, where a 90-year-old water supply conduit cracked at the fringe of UCLA and flooded streets, garages, and sports halls. The period between 2000 and 2010 witnessed 13 spectacular system breakdowns in Polish cities that resulted in the paralysis of almost entire district and led as a consequence to disruptions in the functioning of entire cities that lasted for some time. The breakdowns were related both to failures of water supply networks and to defectiveness of wastewater disposal networks. They were often related to susceptibility to failure that resulted from faulty operation of networks, mechanical damages, errors in network operation, unpredictable and extreme atmospheric conditions (impact of torrential rains on wastewater disposal networks or of severe draughts on water supply networks), as well as from aging of materials, etc. Research conducted in this domain provides a broad perspective on the relations between materials used, types of failures, and reliability analyses, as well as many other factors that influence the functioning of both water supply and wastewater disposal networks.

One of the major large-scale failures occurred in 2002, when 70,000 town inhabitants were deprived access to potable water for about 24 h. The failure was caused by a cracked conduit with a diameter of 1200 mm that supplied water to rapid-flow filters at the water purification plant. The press was informed about this failure event. This allowed the authorities to avoid panic among water consumers. The effects of this situation could have been minimized if a water reserve in tanks had been available during the failure period [14,31].

In 2003, a cracked cast iron pipe (400 mm in diameter) in a zone pumping station was the cause of a water supply network failure in a town of 200,000 inhabitants.

In mid-2003, a town of 50,000 inhabitants suffered an incident of water contamination in the water supply network that lasted for 17 days. Coli group bacteria were detected in the water supply pipeline in the amount of MPN = 99 (99 coli bacteria in 100 cm³ of water, while the admissible standard is 0). Due to information chaos, the town succumbed to panic and inhabitants started to buy mineral water from shops. The contamination was caused by improper operation of the water supply network. The network was not rinsed properly after the application of phosphate preparations used to facilitate removal of deposits from the network. During the crisis period, the local hospital recorded 100 cases of various types of ailments related to the failure event.

Another failure event in 2003 affected a city of about 2 million inhabitants. An arterial conduit of the water supply network suffered a breakdown in the vicinity of the national stadium. Water flooded 1/4 of the stadium surface area. Many cars were flooded. Moreover, about 25% of the inhabitants had no access to water.

In 2005, in a city of 550,000 inhabitants, a delivery collector sewer of 800 mm in diameter (feeding sewage to one of the two sewage-treatment plants) was damaged during drilling operations. As a result of this event, a part of the sewage had to go without treatment, for a short period of time, over an overflow to the point of disposal. The other part, in turn, was directed to a smaller sewage-treatment plant, which resulted in facility overloading [13].

In 2007, a brick collector sewer with pear-shaped cross-section (1.20×1.50 m) collapsed along a section of 50 m. It was the only point of disposal of sewage from 370,000 inhabitants. The failure resulted from improper selection of materials used to construct the collector sewer (constructed in 1956).

A water supply network in a city of 700,000 inhabitants in central Poland, with a water supply network length of about 1430 km, was analyzed in terms of susceptibility to failure. The water supply network was made primarily of cast iron, steel, PVC, PEHD, asbestos cement, and reinforced concrete. The analysis yielded the following damage types (Tables 9.7 and 9.8):

- Corrosion (also as a cause of damages)
- Transversal cracking
- Longitudinal cracking

Table 9.7 Failure Rate as a Function of Conduit Type

Type of Water Supply	Network Length (km)	Number of Failures in the Period 2008-12	Failure Rate (failure/km year)	One Failure on Pipe Length (km)
Arterial conduits	244	68	0.06	17.94
Distribution conduits	1184	1853	0.31	3.19

Table 9.8 Material Type, Number of Failures, and Failure Rate as a Function of the Material Used to Construct a Water Supply Network

Type of Material	Network Length (km)	Number of Failures in the Period 2008-12	Failure Rate (failure/km year)	One Failure on Pipe Length (km)
Cast iron	643	1397	0.43	2.30
Steel	39	46	0.24	4.24
PVC	424	265	0.13	8.00
PE	208	64	0.06	16.25
Asbestos cement	92	134	0.29	3.43
Others (PP, winidur, reinforced concrete)	22	13	0.12	8.46

- Fixture damages
- Mechanical damages
- Material defects (also as a cause of damages)
- Seal failure
- Fractures

Causes

- Design errors
- Poor workmanship
- Material defects
- Dynamic loads: vehicle traffic
- Water hammer
- Corrosive soil
- Corrosive liquids transported
- Temperature fluctuations (external and external)
- Ground instability
- Stray electrical currents
- Fluctuating and excessive pressure
- Incorrect sealing
- Age

It was assumed that the above damages brought about the following effects in the operation of conduits:

- Exclusion of a conduit section from operation
- Disturbances in conduit operation

According to the adopted classification of researched objects, arterial conduits (A) and distribution conduits (D) were distinguished within the water supply network. Within the network, 1921 failures were recorded during the entire observation period. The majority of failures (96%) occurred on distribution conduits, while only 4% affected arterial conduits. The number of failures alone is insufficient to perform a comprehensive assessment; although it is an important piece of general information about susceptibility of a water supply network to failure. The susceptibility of a network is also a function of its length. Therefore, the average unit number of failures per kilometer per year is decidedly a better index for the purposes of assessment. The failure rate amounted to 0.06 for 244 km of arterial conduits and to 0.31 for 1184 km of distribution conduits.

An analysis was performed of types and effects of damages to conduits related the material implemented to manufacture them. A preliminary data analysis showed that damages to conduits made of thermoplastics. (PVC, PE, and polypropylene; PP) and of asbestos cement required mainly the shutting-off of a damaged section, which resulted in discontinuation of water supply to consumers. Due to limited data availability, the relations of types and effects of damages are presented for the conduits that are most susceptible to failure, that is, for cast iron and steel conduits (Tables 9.9 and 9.10).

It follows from the analysis of types and effects of damages to cast iron and steel conduits that:

- Conduits made of steel were primarily affected by corrosion (74% of failures), and only every fourth failure affecting these conduits did not result in a discontinuation of water supply to consumers.
- The highest number of failures was recorded for cast iron conduits. The failures were caused by a wide variety of damages, primarily longitudinal and transversal cracking responsible for the highest number of failures. Every fifth reported failure was caused by forced-out seals, while every seventh failure resulted from a mechanical damage. In 88% of cases, the damages resulted in excluding a conduit from operation.

The analysis of types of failures of other conduits allows one to formulate the following conclusions:

- PVC conduits were affected by mechanical damages (over 3/4 of all failures). To a much lower extent were they affected by transversal and longitudinal cracking and by forcing the sealing material out of pipe sockets. In 99% of cases, the failures affecting PVC conduits resulted in excluding a conduit section from operation.
- PE pipelines were affected by mechanical damages as well as longitudinal and transversal cracking.
- Pipelines made of asbestos cement were affected primarily by damages in the form of transversal and longitudinal cracks (51% and 31%, respectively). Apart from that, cases of forced-out seals and mechanical damages were recorded.

Table 9.9 Relations of Types and Effects of Damages for Cast Iron Conduits

Effect	Failure							Σ (%)
	Corrosion (%)	Longitudinal Crack (%)	Transverse Crack (%)	Valve Damage (%)	Mechanical Damage (%)	Material Defect (%)	Pushing Seals (%)	
Pipe section disconnection	7	17	36	1	13	1	14	88
Work disruption in pipe	1	2	3	0	1	0	4	12
Σ	8	18	40	1	14	1	19	100

Table 9.10 Relations of Types and Effects of Damages for Steel Conduits

Effect	Failure						Σ (%)
	Corrosion (%)	Longitudinal Crack (%)	Mechanical Damage (%)	Material Defect (%)	Pushing Seals (%)		
Pipe section disconnection	59	4	7	2	4		76
Work disruption in pipe	15	0	2	0	7		24
Σ	74	4	9	2	11		100

For the purpose of analyzing susceptibility to failure, a spatial distribution of failures of the water supply network was prepared on the basis of the data collected in the last year under investigation (2012). In order to present the spatial distribution of the degree of susceptibility of the water supply network to failure, a grid of uniform fields with the surface area of 4 km^2 ($2 \times 2 \text{ km}^2$) was formed and failures were counted in each field. Figure 9.9 presents a fragment of the map of the water supply network with the superimposed grid of square fields and the numbers of failures corresponding to each field.

Further conclusions can be formulated on the basis of the failures data in 2012 and the failure distribution maps as well as knowledge of other terrain-related conditions, like the depth of the groundwater layer and terrain geology:

- 1921 failures affecting water supply network conduits were recorded during the entire observation period (2008-12). The majority of failures (96%) affected distribution conduits, while only 4% affected arterial conduits.
- The highest number of damages (73%) affected cast iron conduits. Pipelines made of other materials were affected by damages much less frequently, namely: PVC conduits: 14%, and conduits made of asbestos cement: 7%. Conduits made of other materials were affected by isolated failure events.



FIGURE 9.9

Map of spatial distribution of failures of the water supply network based on the last year of observation (2012).

- During all the years under investigation, the highest numbers of failures were recorded in cold and autumn periods. In general, the highest numbers of failures were recorded in winter, except for 2008, when the highest number of failures was recorded in autumn. The lowest numbers of recorded failures occurred in spring and summer. The above-described failure distribution over the four seasons is clearly related to changes of the air temperature and the correlated changes of the temperature of the ground and the temperature of water transported through the network.
- In 2012, 15 failures were recorded to have affected arterial conduits. Almost all of them affected cast iron conduits. 2/3 of failure events involved forcing out of the sealing material. Arterial conduit failures occurred with similar frequencies in all four seasons. Only one failure event was related to the impact of the surrounding ground, where the groundwater level was less than 2 m. The pipeline in question was constructed in sandy soil.
- In the case of distribution conduits, the majority of failures occurs in winter time. Most often, they affected cast iron conduits that suffered from transversal and longitudinal cracking. Oftentimes, pipelines suffered failures due to mechanical damages or sealing damage. Every fifth damage influenced the surrounding groundwater reaching above the level of network conduit installation. More than half of the damaged conduits were built on boulder clays. Apart from that, water supply conduit failures occurred in sandy soils (25% of all failure events).
- Conduits made of asbestos cement were more often affected by failures in autumn and winter (60% of failure events). They suffered primarily from transversal and longitudinal cracking. Failures of asbestos cement conduits affected conduits laid in clays, sands, and glacial lake deposits.
- In the case of PE conduits, no seasonal correlation of susceptibility to failure was observed. These conduits were affected primarily by mechanical damages. Only one failure event affecting a pipeline located below groundwater level was recorded.
- Two failure events affecting PP conduits were recorded. Both occurred in autumn and were caused by transversal cracking. They did not affect groundwater and the damaged pipelines were installed in clays and glacial lake deposits.
- The seasons most probably have no influence on the vulnerability of PVC conduits. These conduits most often suffered from cracking and 30% of these failure events could affect groundwater or result in secondary contamination of water in the water supply network.
- Conduits made of steel suffered damages both in the summer and in winter. They were usually affected by corrosion.
- The vulnerability of cast iron conduits was decidedly higher in winter months. These conduits usually suffered from transversal and longitudinal cracking. Every sixth failure event affected groundwater, while over a half of the damaged cast iron conduits was installed in boulder clays.

- Seasonal temperature changes mainly influenced the susceptibility to failure of distribution conduits. A clear correlation was observed in the case of cast iron conduits: 70% of failure events occurred in autumn and winter. In winter months, pipelines were damaged primarily due to transversal and longitudinal cracking and to forcing out of seals. In summer months, in turn, they were affected by longitudinal cracking, mechanical damages and transversal cracking. Groundwater was affected by every fifth failure in winter and every sixth failure in the summer.

Irrespective of the season, when a failure of the water supply network occurred:

- Over a half of failure events concerned conduits laid in boulder clays, while every fourth failure of the water supply network affected pipelines laid in fluvioglacial and glacial sands. Other soil types did not show substantial correlation with the susceptibility of water supply conduits to failure.
- Over 80% of failure events affected conduits laid above the groundwater level and only 18% affected conduits operating below groundwater table. This tendency is confirmed for cast iron conduits and for conduits made of PVC.

The above analysis used the criterion for assessing susceptibility to failure presented in [Table 9.5](#). On the basis of performed calculations, it was discovered that arterial conduits are characterized by low susceptibility to failure, while distribution conduits—by medium susceptibility to failure. 0.06 failures per kilometer per year affect the former category of conduits, while the latter category is affected by five times more failures, that is, 0.31 failures per kilometer per year.

In terms of the material used to manufacture conduits, cast iron conduits are characterized by the highest susceptibility to failure amounting to 0.43 failures/km·year. Next in the ranking of susceptibility to failure correlated with the length of pipelines are conduits of asbestos cement and steel conduits. Conduits made of plastics—namely PVC, PE, and PP—are characterized by the lowest susceptibility to failure. According to the adopted classification scheme, only PE conduits fall into the category of conduits with low susceptibility to failure.

The distribution of failures of water supply network as a function of the month of a failure event shows that seasonal variations (of temperature) have an impact on network failure frequency. The highest failure rate was observed in winter months, that is, January, February, March, and April. The lowest number of failures per kilometer was recorded in June.

6 CONCLUSION

The foregoing analysis based on the results acquired during network operations yielded a highly diversified outcome for similar or even identical objects under investigation. Therefore, the most objective approach consists in presenting the range of values describing susceptibility of these materials to failure. In the case

of water supply networks, the lowest degree of susceptibility to failure (irrespective of the number of factors affecting this value) is assigned to transit conduits and some arterial conduits. The corresponding range of values of λ is 0.04–0.1 failure/km·year. Distribution conduits, in turn, are characterized by high susceptibility to failure that oscillates around 1 failure/km·year. When the conduit type is correlated with its function, one gets a correlation between failure rate and conduit diameter. This correlation tends to assume decreasing values as the diameter increases. Irrespective of the actual age and diameter of conduits, an increase of susceptibility to failure is also correlated with changes of the temperature of the ground and water (or sewage, in case of wastewater disposal systems) and to changes of air temperature. On the other hand, the failure rate drops, when the pressure in a water supply network is reduced. An example based on the data from Wrocław: pressure reduction by 40% resulted in the reduction of failure rate by about 41%, while pressure reduction by 20% in Oleśnica resulted in the reduction of the failure rates of conduits made of gray cast iron by 53% and by 63% in the case of PVC conduits.

In terms of reliability, conduits made of thermoplastics (PE and PVC) have the best results. They are characterized by the lowest failure rate, compared to conduits made of steel and grey cast iron. It is worth underscoring that the investigated failure rates of PE conduits in Poland—falling within the range between 0.02 and 0.69 failure/km·year—describes the first generation of these pipelines, installed in the period 1981–90. In other countries, the failure rate for pipelines made of the same material falls within the range of 0.02–0.18 failure/km·year, with very similar values for PVC conduits.

The assessment the susceptibility of water supply and wastewater disposal systems to failure was based on data relating to the vulnerability of particular elements operating in real systems. The level of the system susceptibility to failure is acceptable.

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A reliable analysis method for estimating large excavator structural strength

10

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1 INTRODUCTION

Large excavators such as the mining dragline (Figure 10.1a) and shovel (Figure 10.1b) are used for surface mining to dig and transport overburden [1–10]. Their large size front-end structures can swing either left or right during working. The repeated loading induces the front-end to undergo fatigue loading as a result of significant cracking at welded connections [10,11] and fatigue failure as simulated in Figure 10.2. It is estimated that 90% of structural failures in large excavators are due to fatigue alone. Several decades ago, the boom of Peabody Coal Company's Marion 8750 failed during operation at the Black Mesa Mine



(a)

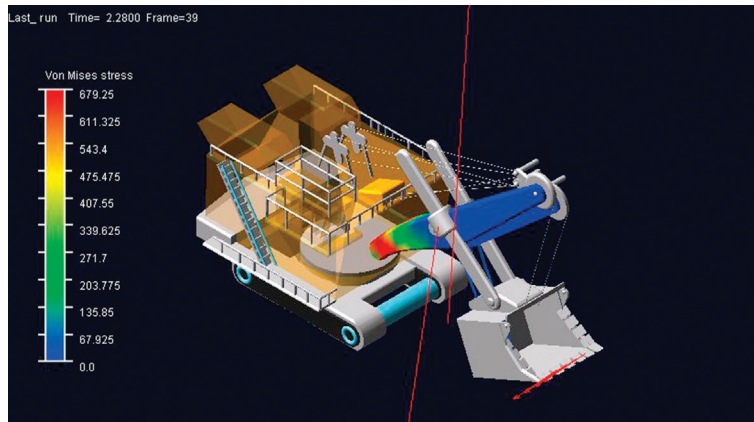


(b)

FIGURE 10.1

Large excavators used for surface mining: (a) mining dragline and (b) mining shovel.

(www.miningmayhem.com). The boom failure was caused by the dynamic stress loading coming through the tight drag and hoist ropes. In recent years, seven large shovel and three walking draglines had experienced some boom or tri-structure lacing cracking not long after going into service, which resulted in the maintenance casts ranged between \$20 and \$130 million [1–12]. Therefore, it is important to ensure the excavator longevity and reduce the high capital investment. Researchers had tried to find

**FIGURE 10.2**

Simulated shovel boom failure.

efficient analysis technologies to monitor the overall stress loading of front-end structure to avoid fatigue failure.

The typical analysis methods used for mining excavator structural strength are the field test, static analysis, and dynamic analysis [10–16]. Yin et al. [12] have analyzed the fatigue life of cable shovel boom by employing field test data and finite element analysis (FEA). Some researchers have identified the field test and static FEA as valuable tools in fatigue evaluation of dragline boom. Using measured boom stress data, Dayawansa et al. [13] have studied how the cracks grow in the weld joints in order to protect booms from future catastrophic collapse. Using strain gage measurements and static FEA model, Joshi et al. [11] have analyzed boom hot spot stresses to monitor structural health. For the front-end structure, Li and Liu [14] have developed a dynamic flexible model to detect component fatigue failure using multi-body dynamics and the FEA method. This research demonstrates that the dynamic analysis method can be less costly and more time efficient than the field test method, and also can reflect the dynamic character of a structure, which is missed and not reflected from static analysis. However, the research is focused on system dynamics modeling. The structural dynamics modeling of dragline front-end was still in the preliminary stage and the model validation was not addressed, which is necessary to verify the simulation results.

The objective of this research is to apply the latest technological advances of dynamic analysis to predict the large excavator fatigue life. As a case study, cable shovel front-end strength analysis is addressed. First, the structural components are described and the operation process is introduced. Second, a front-end flexible model is created to calculate component stress-time histories by inputting a working cycle loading. Third, the dynamic model is simulated and validated by comparing the stress range (SR) with static analysis results and field testing results. Fourth, the structural fatigue lives at hot spots are evaluated.

2 SHOVEL STRUCTURE AND OPERATION PROCESS

2.1 STRUCTURE DESCRIPTION

Figure 10.3 shows a mining shovel diagram with main structural members, which include the base frame, revolving frame, and front-end mechanism. The base frame serves as a stable base for the shovel. The revolving frame functions as a platform for front-end attachment, machinery house, and operator's cab. The front-end mechanism consists of a boom, handle, and dipper. The dipper is suspended from the boom with wire ropes and is maneuvered in hoist-crowd-swing directions.

2.2 OPERATION DESCRIPTION

Figure 10.4 demonstrates the work cycle of the shovel which comprises eight distinct phases [15]: (1) dropping phase in Figure 10.4a, where the dipper is positioned above the material to be excavated; (2) digging phase in Figure 10.4b, where the dipper is lowered and the crowd handle is drawn so that the dipper is crowded along the surface of the material; (3) lifting phase in Figure 10.4c, where the dipper is lifted by using the hoist rope; (4) swinging accelerating phase in Figure 10.4d, where the front-end is swung forward fast; (5) swinging decelerating phase in Figure 10.4e, where the front-end slows down to where the material is to be dumped; (6) dumping phase in Figure 10.4f, where the dipper door is opened and dipper is emptied by the dump operation; (7) swinging back phase in Figure 10.4h, where the dipper is swung

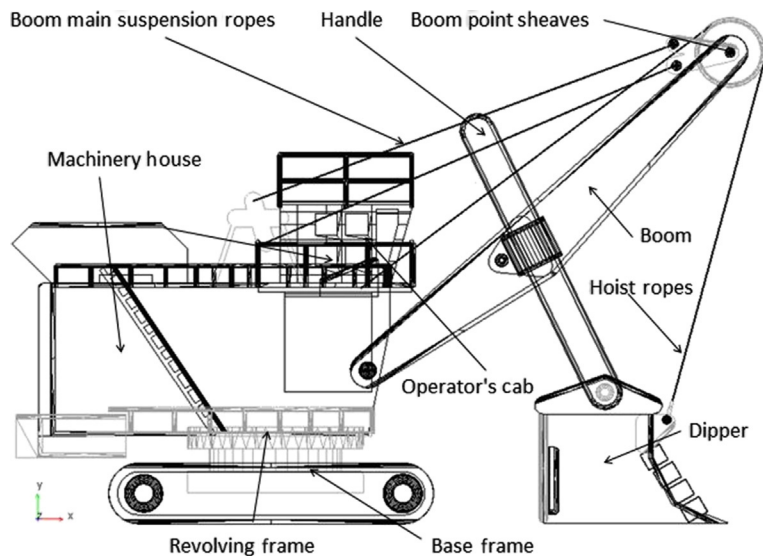


FIGURE 10.3

A mining shovel with main structural members.

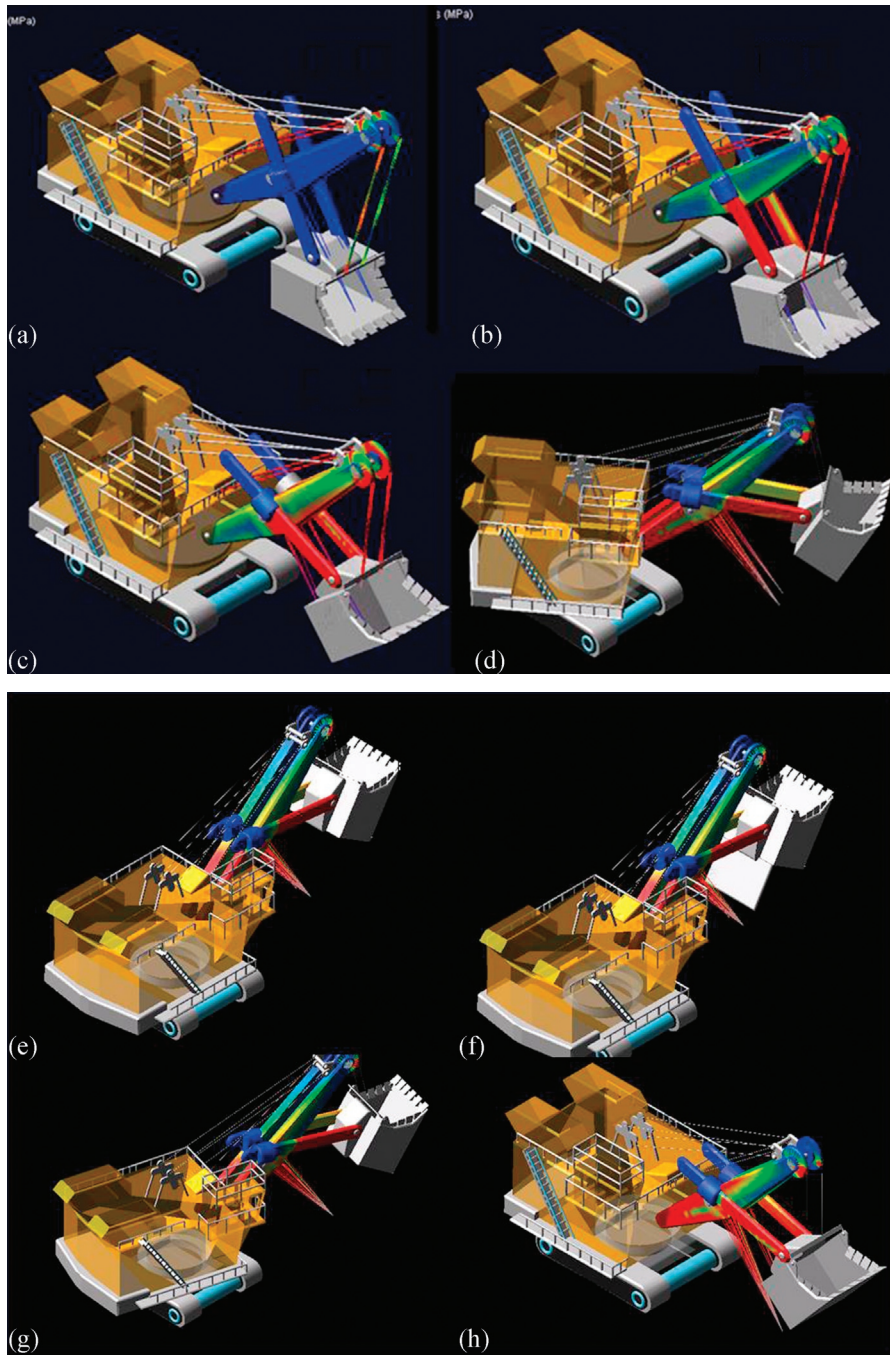


FIGURE 10.4

A typical shovel operation cycle illustrated in eight phases: (a) dropping phase, (b) digging phase, (c) lifting phase, (d) swinging accelerating phase, (e) swinging decelerating phase, (f) dumping phase, (g) swinging back phase, and (h) returning phase.

in the reverse direction; and (8) returning phase in Figure 10.4g, where the dipper is returned to the pit and will repeat the cycle. One loading cycle generally takes 40-60 s during normal operations [3].

3 STRUCTURAL DYNAMICS ANALYSIS METHOD OF SHOVEL FRONT-END MECHANISM

The structural dynamics model of the shovel front-end system with all the external loadings and accelerations is created using finite element method (FEM). FEM is a powerful technique to apply the numerical method to solve complex engineering structural problems. In the FEM, the excavator front-end components are modeled as the series of finite elements interconnected by nodes with the physical properties such as elasticity modulus, Poisson's ratio, density, yield strength, and ultimate tensile. For the dynamic finite element stress problem, transient dynamic analysis is proposed to determine the stress-time histories of components. And equivalent SRs are calculated. Figure 10.5 describes the three areas for the modeling, loading, and stress analysis. The details of each area are discussed below.

3.1 FINITE ELEMENT MODELING OF FRONT-END STRUCTURE

Figure 10.6 presents the shovel front-end nonlinear FE model, where the global coordinate O (X,Y,Z) is fixed to the ground at the revolving frame center. The front-end parts are built as flexible multi-bodies, which are the boom and handle with solid

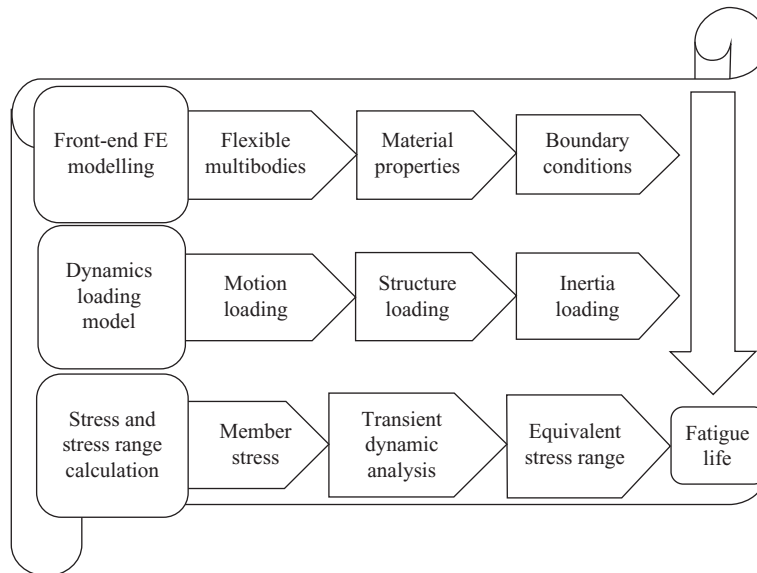


FIGURE 10.5

Structural dynamics modeling and analysis procedure of shovel front-end.

elements, the suspension ropes with tension cable elements, the rigid revolving frame with shell elements, and the others with node mass elements. The model does not include the flexible bucket because the loads generated from three parts are calculated using a shovel rigid multi-body dynamics model developed by Li and Frimpong [16]. Here, the loads are applied to hoist rope.

The material properties of the components consist of the elasticity modulus of 206.7 GPa, Poisson's ratio of 0.3, density of 7820 kg/m³, yield strength of 150 MPa and ultimate tensile of 290 MPa. The modal mass, stiffness, and deflection characteristics are calculated by employing the FEM. A damping ratio of 2-4% is assigned to modal frequencies between 1 and 20 Hz.

Referring to the shovel operation described in Section 2.2, the model is allowed to rotate about the y direction at the center O and other five degrees of freedom are restrained, which are the x, y, and z axial directions, and x and z rotation directions. The kinematic constraints or forces are applied to boundary nodes to connect two parts. The motion of flexible bodies is defined by Lagrange equation 10.1 [17].

$$M\ddot{\xi} + \dot{M}\dot{\xi} - \frac{1}{2} \left[\frac{\partial M}{\partial \xi} \dot{\xi} \right]^T \dot{\xi} + K\xi + \frac{\partial V_g}{\partial \xi} + D\dot{\xi} + \Psi_{\xi}^T \lambda = Q \quad (10.1)$$

3.2 DYNAMICS LOADING OF FRONT-END STRUCTURE

Figure 10.6 also contains the local coordinate o (x,y,z) which is attached to the revolving frame at the center O for loading purposes. The model loading has been discussed in our early publication [15], which is summarized as:

- Motion loads going into the front-end with the digging, hoisting, swinging, dumping, and swinging back;
- Structural loads at the hoist ropes;
- Inertia loads due to swing acceleration.

3.3 EQUIVALENT SR AND LIFE ESTIMATION OF FRONT-END STRUCTURE

The fatigue life of the front-end structure is predicted by employing Palmgren-Miner's rule along with a Rainflow cycle-counting procedure. The stress-time history of each structural element is obtained using transient dynamic analysis. Figure 10.7 shows a hot spot stress-time history for the front-end structure. It is clear that SR is variable in amplitude. The equivalent SR used to replace the variable amplitude SR is required for fatigue investigation, because experimental fatigue test is built from constant amplitude stress. For getting variable amplitude fatigue, all recovered structural member stresses are first processed with Rainflow [18] for cycle-counting the SR of the variable amplitude. The irregular stress-time history is decomposed into equivalent stress columns. The number of cycles in each column is recorded in a SR histogram as shown in Figure 10.8.

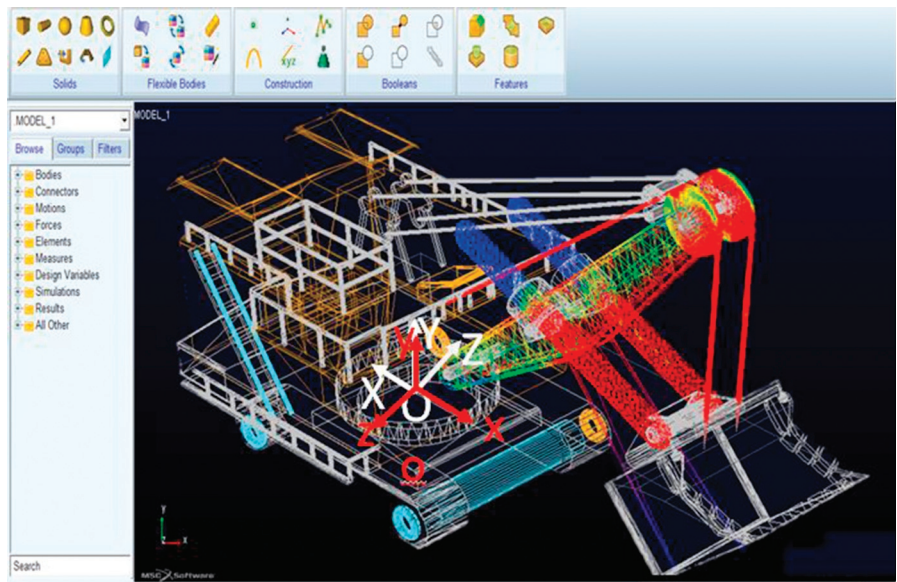


FIGURE 10.6

Shovel front-end flexible model.

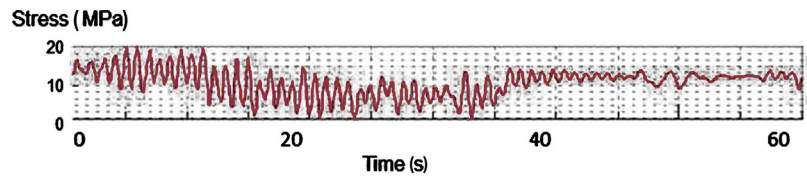


FIGURE 10.7

Hot spot stress-time history of structural component.

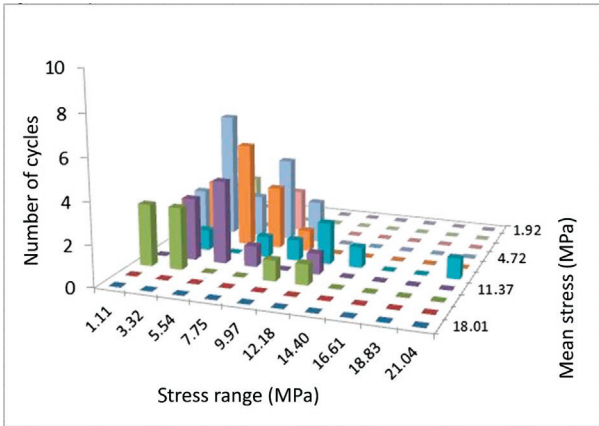


FIGURE 10.8

Rainflow matrix histogram of hot spot stress.

The damage summation from each cycle needs to be performed to obtain the total amount of fatigue damage due to all cycles. Palmgren-Miner's rule is proposed to calculate the structural fatigue damage for the mining excavators [19]. This theory is described by the S-N curve as given in Figure 10.9. If there are k different SRs or levels and n_i cycles are applied at SR (where the fatigue life is N_i cycles), then Palmgren-Miner [20] gives the cumulative damage amount D as Equation 10.2.

$$\sum_{i=1}^k \frac{n_i}{N_i} = D \quad (10.2)$$

In this study, D at an equivalent constant amplitude SR σ_v is linearly proportional to the ratio of the experienced number of cycles n_v to the number of cycles to failure N_v at this stress level. The above equation can be expressed as Equation 10.3.

$$\sum_{i=1}^k \frac{\sigma_i^m n_i}{\sigma_v^m N_i} = \frac{\sigma_v^m n_v}{\sigma_v^m N_v} \quad (10.3)$$

where m is the crack growth rate constant. From Equation 10.4:

$$\sigma_i^m N_i = \text{Constant} \quad (10.4)$$

The Equation 10.3 can be written as Equation 10.5.

$$\sum_{i=1}^k \sigma_i^m n_i = \sigma_v^m n_v \quad (10.5)$$

A variable amplitude stress spectrum is replaced by an equivalent constant amplitude SR by Equation 10.6. Using this SR, the fatigue life for this hot spot is determined from the S-N diagram for the given structure material.

$$\sigma_v = \left(\sum_{i=1}^n (\sigma_i^m n_i) / n_v \right)^{1/m} \quad (10.6)$$

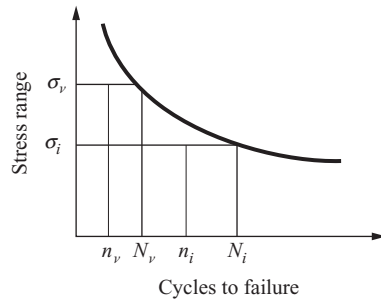


FIGURE 10.9

S-N curve diagram described by Palmgren-Miner rule.

4 RESULTS AND DISCUSSION

A heavy shovel is employed to simulate and validate the model against static and test data. The fatigue lives of front-end hot spots are then estimated. The structural dynamics of the shovel is simulated using the hoist rope force and swing motion as inputs. The dynamic stresses of front-end structural hot spots are recovered and compared with the static and test results reported by Zhang [21], where (i) the static stresses of boom and handle were calculated for the shovel; and (ii) the dynamic stresses of boom and handle were investigated by analyzing field test data on the shovel. The details are presented below.

4.1 SIMULATION AND VALIDATION OF FRONT-END MODEL

Figure 10.10a and b indicates the dynamic loading on the front-end model in the local coordinate system $o(x,y,z)$ during the one working cycle of 60 s. They consist of Figure 10.10a with the load at the hoist rope; and Figure 10.10b, the swing angular displacement at the revolving center o in the y direction.

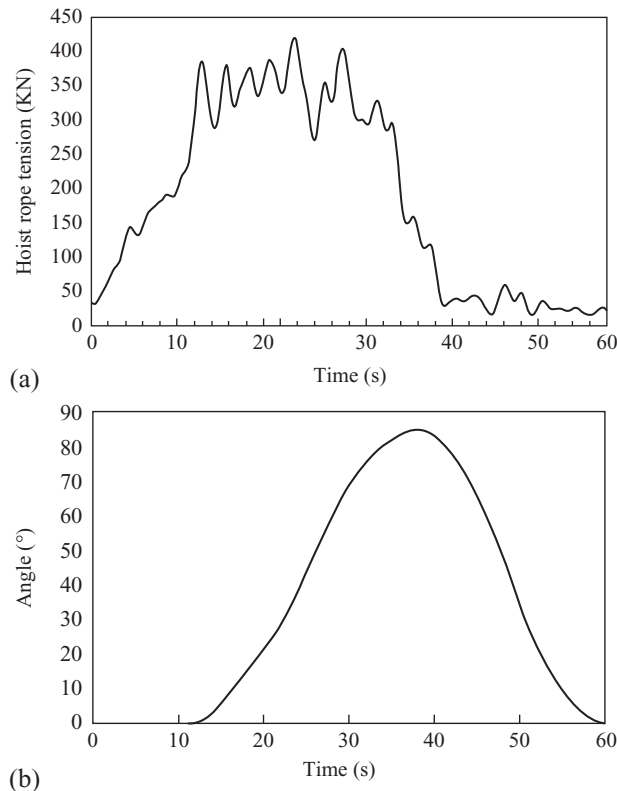


FIGURE 10.10

Dynamic loads on the shovel front-end with time: (a) load at hoist rope and (b) swing angular displacement at the revolving center.

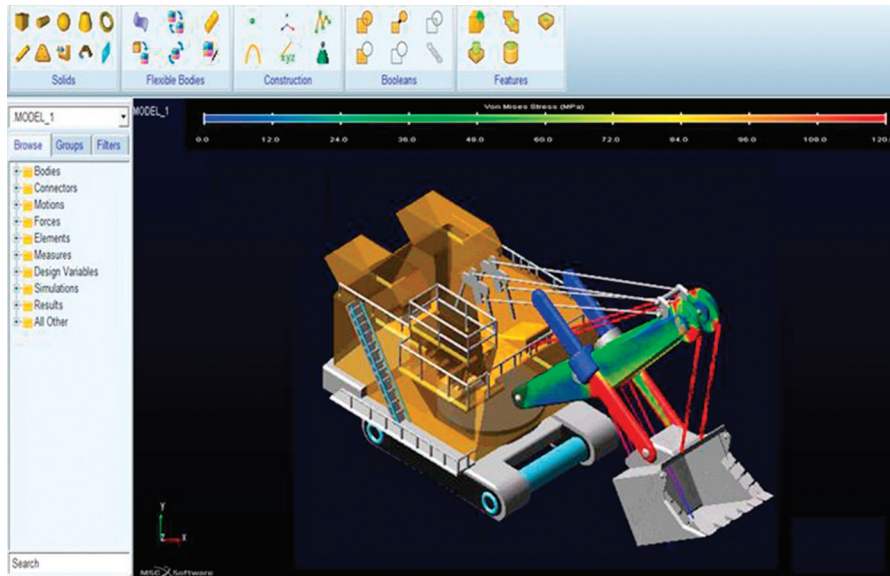


FIGURE 10.11

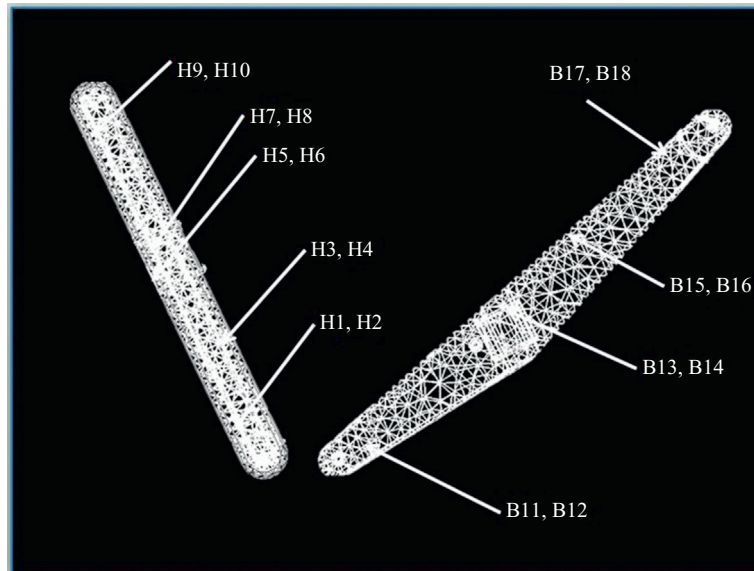
Shovel transient dynamic simulation for one operation cycle.

The dynamic stresses of front-end hot spots are obtained by transient dynamic simulation for one operation cycle as given in [Figure 10.11](#). The 18 hot spots that are of interest in this study case are indicated in [Figure 10.12](#). [Figure 10.13a](#) indicates the dynamic analysis stress range (DASR), the static analysis stress range (SASR) [21], and test analysis stress range (TASR) [21] for the 18 hot spot stresses (see [Figure 10.12](#)). It can be seen that for each hot spot, the DASR is bigger than SASR or TASR. However, for the boom hot spots, the SASR is greater than the TASR; and for the handle hot spots, the SASR is smaller than the TASR. The detailed comparison of three results is presented in [Figure 10.13b](#) and described below.

Comparing the DASR with SASR results, the minimum ratio (DASR/SASR) of 1.29 occurs at the handle hot spot 3 and the maximum one of 3.07 does at the handle hot spot 8. The ratios of all hot spots are >1 , which indicates that the SR of 18 hot spots is bigger in the dynamic analysis than in the static analysis.

Comparing the SASR with TASR results, the static ratio (SASR/TASR) is larger than 1 for the boom hot spots and <1 for the handle hot spots 7-10. That means the SASR of some handle hot spots is smaller than the test results. The result indicates that the static analysis overestimates the SR in the booms, but underestimates the SR in the handle hot spots.

Comparing the DASR with TASR results, the dynamic ratios (DASR/TASR) for the 18 hot spots are between 1 and 2.5. The minimum ratio of 1.05 appears at the handle hot spot 10 and the maximum one of 2.45 does at the handle hot spot 2. That means the DASR for the hot spots is significantly bigger than the TASR result. The difference

**FIGURE 10.12**

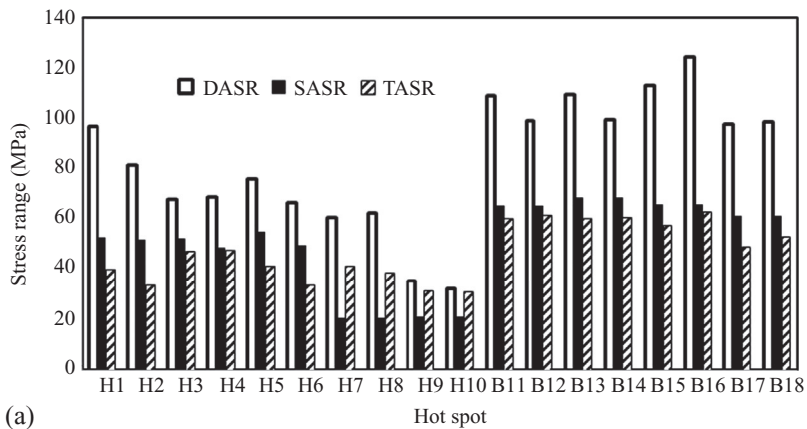
Hot spot locations at handle and boom.

between the DASP and TASR results is attributable to the inertias in the motor, gear-box, drum, and rope, which were not taken into account in the test stress analysis [22].

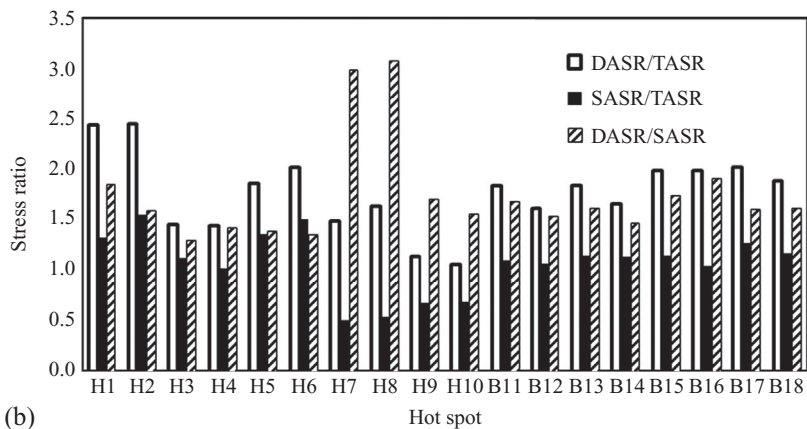
This analysis case validates the potential of the dynamic model to predict the front-end structural strength. The analysis results identify that the dynamic analysis will give a higher safety confidence in the shovel front-end stress simulation than the static analysis. The analysis results also illustrate that the dynamic shovel model can capture the dynamic motion of front-end structure under normal digging operation conditions. The differences between the model prediction and test data can be attributed to the indirect forms of test, the simplifications of FEA front-end model, and loading errors.

4.2 SHOVEL FRONT-END FATIGUE LIFE PREDICTION

The shovel operates 2300 cycles per day [12]. The front-end life prediction in years is calculated based on the SR result for each hot spot. Figure 10.14 displays the fatigue life estimation of the 16 hot spot stresses. The hot spots of H9 and H10 are not included in this estimation because their fatigue lives are infinite. The fatigue lives of handle hot spots vary from 8.47 to 54.69 years with an average of 32.31 years; and the fatigue lives of the boom hot spots vary from 3.06 to 8.16 years with an average of 6.12 years. It is clear that boom hot spots have lower fatigue lives than handle hot spots. The minimum value of 3.06 years is at boom hot spot 16 and the maximum of 54.69 years is at handle hot spot 7.



(a)



(b)

FIGURE 10.13

Shovel front-end stress range results: (a) dynamic analysis, test analysis, and static analysis; and (b) comparison between the dynamic analysis, test analysis, and static analysis.

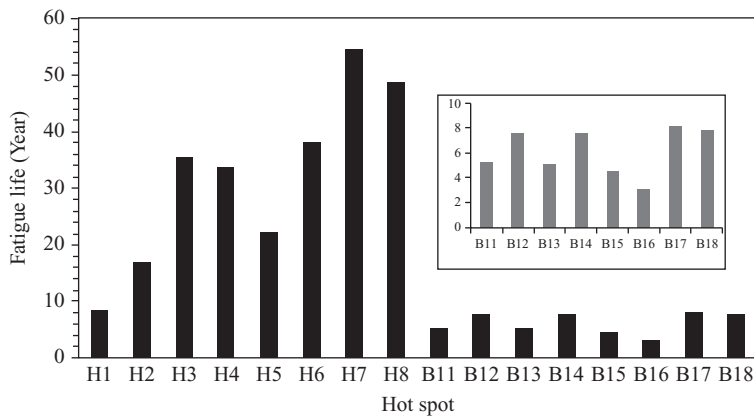


FIGURE 10.14

Predicted fatigue life for shovel front-end hot spots.

5 CONCLUSION

A comprehensive approach to predict the structural strength of large excavators is introduced. The method is validated by employing a heavy mining shovel. The shovel front-end flexible FEA model is created. The loads applied on the model are determined from the external loading and acceleration. The stress-time history of each structural element is obtained using transient dynamic analysis. The case analyses are presented to validate the method, where comparisons of equivalent SRs between the dynamic analysis with the static analysis or field test are conducted. The fatigue lives at hot spots are predicted.

It can be concluded that the dynamic analysis gives a more economic, reliable, and safety confidence in the shovel front-end stress simulation. The front-end structure has created a flexible multi-body virtual prototype. The dynamic stress-time history has been captured during the simulating virtual reality shovel operation. Comparing the methods between the virtual prototype analysis and the physical prototype analysis, it can be seen that the former spends less money and time than the latter. Comparing the hot spot SRs from dynamic analysis and static analysis, it can be seen that the dynamic character is missed and not reflected from the static analysis. Comparing the hot spot SRs from dynamic analysis and field test, it can be seen that the former gives larger SR than the latter. With the validated model, the fatigue lives at hot spots are calculated, which is useful to determine design change in the front structure.

This research provides an efficient method for the investigation of mining equipment structural strength. Future research will create submodels for further structural fatigue life evaluation.

NOMENCLATURE

D	damping
DASR	dynamic analysis stress range
λ	Lagrange multipliers
FEA	finite element analysis
$\hat{i}$	generalized coordinates
K	generalized stiffness matrix
m	crack growth rate constant
M	generalized mass matrix
n_i	number of cycles accumulated at stress range σ_i
N_i	number of stress cycles to failure at stress range σ_i
n_v	equivalent number of stress cycles at equivalent stress range σ_v
N_v	equivalent number of stress cycles to failure at equivalent stress range σ_v
ξ	kinematic constraint equations applied to the flexible body
Q	generalized applied forces
SASR	static analysis stress range

SR	stress range
TASR	test analysis stress range
V_g	gravitational energy
σ_i	i th stress range in the spectrum
σ_v	equivalent stress range

ACKNOWLEDGMENTS

This work was supported by the senior author's family members. I thank them for bringing encouragement to my life. Years ago, I had experienced extreme fatigue due to a health problem. I was unable to use the computer due to extreme muscle pain. I was discouraged and hopeless. However, my son bought a Kindle for me so that I was able to read and my husband bought me a tablet so that I could continue to type. I told myself to be strong in this time of hardship. Day by day, month by month, and year by year, I worked on and finished my manuscript which I enjoy doing. I am glad to have contributed my knowledge to the scientific community.

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The reliability design of mechanical system and its parametric accelerated life testing

11

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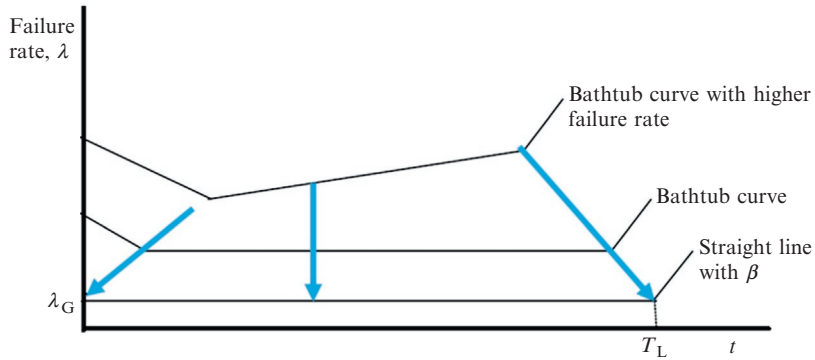
CHAPTER OUTLINE

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1 INTRODUCTION

For mechanical systems, robust design techniques [1–4] were developed by statisticians many years ago. Robust design processes include concept design, parameter design, and tolerance design. Taguchi's robust design method uses parameter design to place the design in a position where random "noise" does not cause failure and to determine the proper design parameters and their levels. The basic idea of parameter design in the Taguchi's robust design is to identify appropriate settings of control factors that make the system's performance robust in relation to changes in the noise factors. Thus, the control factors are assigned to an inner array in an orthogonal array, and the noise factors are assigned to an outer array.

However, a large number of experimental trials in the Taguchi product array may be required because the noise array is repeated for every row in the control array. For a simple mechanical structure, a lot of design parameters should be considered in the Taguchi method's robust.

**FIGURE 11.1**

Bathtub curve and straight line.

Reliability refers to the ability of system or component to perform a required function under stated environmental and operational conditions for a specified period of time [5]. Traditionally, the reliability over the product life can be illustrated by a bathtub curve that has three regions: a decreasing failure rate, a constant failure rate, and an increasing failure rate, as shown in Figure 11.1.

As the reliability of a product (or part) improves, failure rate of the product declines in the field. The bathtub with higher failure rate may be changed into a straight line with shape parameter β .

To find out the product reliability (or failure) in a straight line with β , product lifetime L_B and failure rate λ might be quantified in Equation 11.1:

$$F(L_B) = 1 - R(L_B) = 1 - e^{-\lambda L_B} \cong \lambda L_B \quad (11.1)$$

where R is reliability function, λ is failure rate, and t is use time.

Today myriad innovations are constantly emerging through advanced technology. Products based on these innovations improve our lives and widen their boundaries. But new products also experience frequent malfunctions that should have been resolved before they were released into the global marketplace.

Recently, the Boeing 787 Dreamliner experienced some problems due to new design elements—the fuselage of carbon-fiber-reinforced plastic (CFRP) and the electrical system incorporating lithium-ion batteries, which resulted in grounding. Such major issues are ultimately the responsibility of the Chief Executive Officer of the business that launches the imperfect product. CEO must act in advance of the release of new designs. But how can CEO understand concepts generally consigned to engineers? Let's consider the concepts of product verification to find out malfunctions in a new product, which might be the process CEO might ensure before new product release.

2 DESIGN VERIFICATION ACTIVITIES IN R&D, QA (PERFORMANCE/RELIABILITY), AND MANUFACTURE QC AND THEIR TRAPS

Completing the design of a new product requires two kinds of activities—managerial and technical skills. Managerial skill includes adopting a process improvement approach, such as Capability Maturity Model Integration (CMMI), and controlling quality, which Japan has pursued for over 60 years (starting with training by Dr. W. Edwards Deming at the Japanese Union of Scientists and Engineers) [6]. Technical skill involves using a product-specific validation and verification approach.

CMMI has been developed by the Software Engineering Institute at Carnegie Mellon University [7]. It outlines five organizational levels; from lowest to highest, these are initial, managed, defined, quantitatively managed, and optimizing. The purpose of assessing the level of organization is to raise it to the highest level, at which CEOs expect to produce perfect products. This is not an exact approach, but a technique to back up and assess the principal manufacturing process.

Quality control is mainly related to manufacturing. Its focus is how to assure that item variations are within the tolerances of already determined specifications. Therefore, quality control methods are dimensionally different than the verification of new product designs, since the product developer should determine and establish the necessary specifications for new products as well as their tolerances. Quality control is not generally an activity in the design area, but a necessary activity in the manufacturing field in Figure 11.2.

Let's consider product-specific verification as a technical skill. Generally, engineers check numerous design items when developing new products. In the NASA Systems

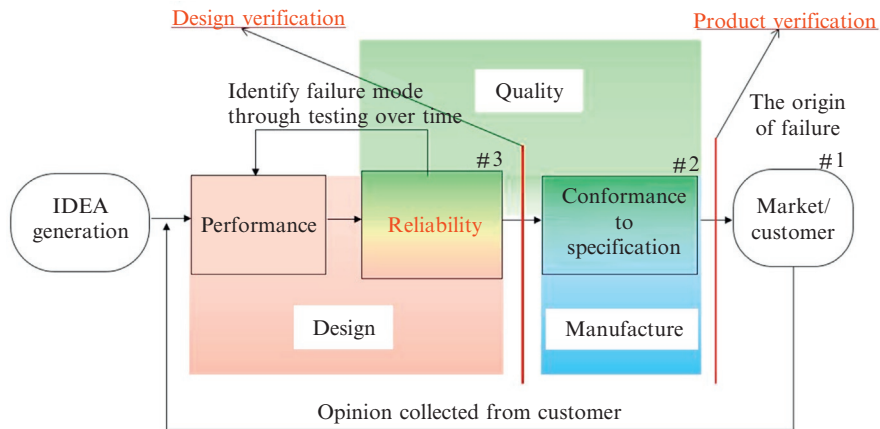


FIGURE 11.2

Confusing quality control: (1) R&D, (2) QA (performance/reliability), (3) manufacture QC.

Engineering Handbook [1], there are 16 activities under the heading “Verification Procedures,” almost all of which involve testing. The keywords include identification of test configuration, test objective, test criteria, test equipment, and location of test activity. Similarly, verification of software includes test strategy, test plan, test procedure, test scenario sorting deficiencies, and so on. But these are general comments or recommendations that may vary according to the activity and the test article, and therefore are not mandatory. Of course the test is required if applicable specifications exist, but that is not sufficient.

From a verification viewpoint, NASA Handbook [8] addresses tasks used to test products, but does not establish the detailed specification standards as the frame of reference in which these tasks might be carried out. It is not acceptable for verifiers to use their discretion when verifying product performance. Carefully established specifications prevent verification activities to deviate from the determined process. Thus, when failure occurs, it is possible to determine whether the specifications are inappropriate or whether verifiers are incorrectly conforming to the specifications. Sometimes, we can also identify omissions in the verification specifications.

Verification specifications should be established over the full range of functions fitted to each product. A thorough use of available technology and related measures to address issues might be applied at an early stage of product development. Why don't CEOs require verification specifications for each product? The reason lies in the engineers' answers. Product-assurance specialists may insist that all related verification activities are included in a “thick document.” And they may add that the activities performed are completely reviewed and revised by related specialists. Furthermore, if CEO figures out the technical details, he would understand all he wanted to know (Figure 11.3). This is a kind of trap.

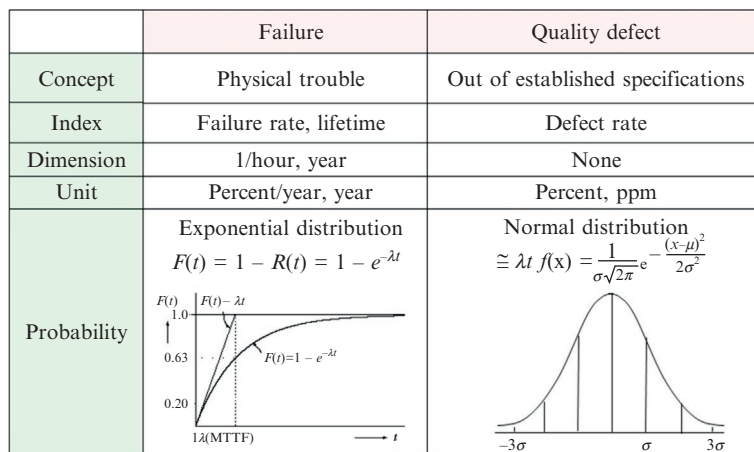


FIGURE 11.3

Differences between quality defects and failures.

3 RELIABILITY QUANTITATIVE SPECIFICATIONS

Technology concepts related to product could be explained with common sense, although some concepts of new technology take time to be understood. Everyone understands new product design concepts if the related staff explains them in every-day communication. Fortunately, the technology concepts related to design verification is not difficult to grasp because we, as consumers, use product. For example, the technology concepts to reduce the noise level of a car engine would be difficult to understand but assessing the performance improvement would be easy since we can hear it. The concepts of design verification-related specification are less complex than that of design itself, and should be easily grasped by CEO. If CEO has difficulty understanding the verification document, there are illogical sections not to be explained with common sense when the engineers write it. The necessary logic in the “thick document” is clarified for any layman to be understood and controversial as seen in Figure 11.4.

Three kinds of simple logic might be addressed in the verification document. First, the information is divided into two activities: verification-specifications establishment and related execution. Second, procedures for how to extract anticipated issues in a new product need to be addressed in the verification specifications, avoiding omissions of necessary specifications and adding some details pertaining to the product for a complete set of specs. Third, verification specifications should be classified into categories according to technological fields in order for related specialists to review their accuracy. Verification specifications might be clearly presented, providing brief summaries to clarify the entire specification concepts.

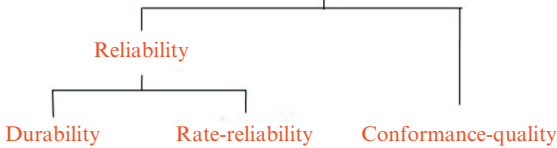
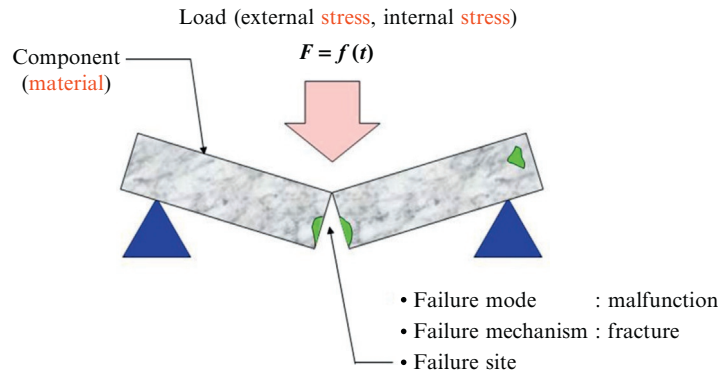
Meaning of basic quality	Product value = Function of product quality and price 		
Concept	Product life	Failure rate	Conformance to specifications
Unit	Year	Percent / year Percent / hour	Percent ppm
Probability	Weibull distribution	Exponential distribution	Normal distribution
Activities	Design change or completion of specification		Inspection, screening

FIGURE 11.4

Technology concepts related to design verification: Double ambiguity of product quality.

**FIGURE 11.5**

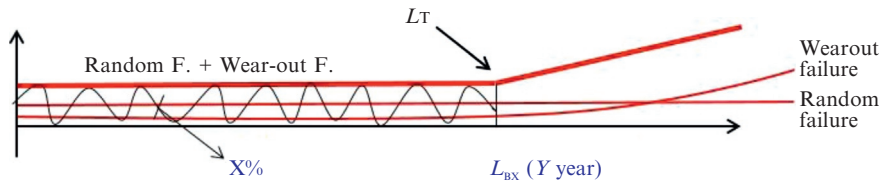
Two elements of failure mechanics, stress, and materials.

Additionally, there are two other issues involved in establishing verification specifications. First of all, we think that new products can check the combinations of specifications used for similar products. This is a misunderstanding. Potential problems inherent in new products cannot be identified using old specifications. The new product incorporates innovative structures, new materials, and different software for upgrading performance and decreasing cost. These cannot be adequately tested using existing specifications. In terms of reliability, adopting new structures may alter failure mechanisms or introduce new ones, as two elements of failure mechanics in Figure 11.5, stress and materials, have changed.

By using the previous specifications, new failure mechanisms are not easy to be identified for products that have the design modifications. In addition to updating the specifications, we should also consider what new testing might be effective. For example, is it possible to apply the test specifications for the Boeing 777 fuselage made of aluminum alloys than to the Boeing 787 Dreamliner fuselage, which incorporates new materials, like CFRP? Obviously, we know that the previous test specs would be improper.

The other issue is that reliability quantitative specifications include estimating item lifetime. Accident caused by material degradation during use could tarnish the company's reputation. But most people consider this task beyond the scope of possibility. Generally, obtaining quantitative results in reliability analysis is very difficult. Reliability specialist Patrick O'Connor wrote in Practical Reliability Engineering that there are basically three kinds of situations—small, moderately large, and very large number of factors and interactions [9]. A small number of factors can be predicted with the physical model. A very large number can be predicted with statistical models. Predictive power diminishes, however, in the case of a moderately large number of factors pertaining to reliability.

Reliability prediction is a necessary task to be undertaken. Let's look at a product in the standpoint of reliability problems. We know that there are a few sites in

**FIGURE 11.6**

Reliability Index; B_X Life (L_{BX}).

product that are weaker than other sites. Reliability specialists can presume the location of the weakest site and/or its failure mechanisms, though they don't know whether the failure will actually happen in the targeted lifetime, or how high the failure rate would be. So if we extract one or two failure sites in the product, mostly in a given module or unit, and classify their failure mechanisms into two categories—lifetime L_B and failure rate λ within lifetime—the factors related to reliability estimation are decreased, and the cases pertaining to moderately large factors become small-factor-number cases. Thus, we can make quantitative estimations about reliability issues—mainly lifetime under normal conditions. This is the simple explanation to understand CEOs.

Let's describe in common sense terms the basic concepts of the required statistics and methods pertaining to establish the quantitative lifetime specification, which CEOs can easily understand as seen in Figure 11.6. For instance, take automobiles. Assume that we test 100 cars in Germany for 10 years and find no trouble (10 years, 160,000 km). We can conclude that the car's failure rate is below 1% per 10 years, which is called "B₁ life 10 years," using a common sense level of confidence. When we conclude that the car's failure rate is below 1% per 10 years, its confidence level reaches around 60%, called the common sense level of confidence. Of course, we cannot test products for 10 years before market release. So we make the accelerated vehicle testing by imposing heavy loads and high temperature until we reach an acceleration factor (AF) of 10. This will reduce the test period by one-tenth, or 1 year. Thus, we test 100 items for 1 year (16,000 km), or 1 week without stoppage (7 days $\times$ 24 h $\times$ 100 km/h = 16,800 km). The next step is to reduce the sample size.

Then, if you increase the testing time, the items would achieve a sufficiently degraded state and many would fail after the test; therefore, we can greatly reduce the sample size, because one or two failed samples would yield enough data to identify the problem area and make corrective action plans. Increasing the test time by four times, or to 1 month, reduces the necessary sample size by the square of the inverse of the test-time multiplier, to one-sixteenth (square of one-quarter), or six engines. The final test specification, then, is that six engines should be tested for 1 month under elevated load and temperature conditions with the criterion that no failure is found. This concept, called parametric accelerated life testing (parametric ALT), is the key to reliability quantitative estimations.

We also cannot guarantee the behavior of a product over 10 years under the extreme environments. These test conditions would be appropriate to mechanical items, like power engines, but the test conditions are not fit to assess the degradation of paint on the automobile body. In addition to testing the new engine, we should devise quantitative test methods for other items—new electrical components (including batteries), electronic control units, lighting systems, or coating materials. According to the identified failure mechanisms, testing must be conducted by subassembly to heighten acceleration.

Without such quantitative lifetime testing, we can't identify all the failures influencing the product's lifetime because there would be unanticipated failures. For example, in my experience of prototype testing, the lifetime of a tub in a washing machine was lengthened at first by small structural changes—a corner radius increase, rib insertion, and so on. The final test, however, showed a weakening of strength in the plastic due to a chemical reaction; the problem was solved by changing the release agent of the injection molding process—something no one could have been predicted as the solution. Note that this method reveals exact failure modes, including totally unexpected ones, that other methods, like failure mode and effects analysis, cannot identify.

For the CFRP of the 787 Dreamliner, the failure mechanism is mainly delamination, which can be found in pressure/humidity/temperature cycling and ultra-violet irradiation testing. If the AF for testing this is calculated using an adequate life-stress model (time-to-failure model) and the sample size is determined according to the BX life target, then quantitative results can be derived. Note that we should check the possibility of failure due to various overstresses, such as bird strikes, with sufficiently degraded samples. For the electrical systems in the 787, engineers should incorporate the same components used in other commercial airplanes (and different combinations of them), assessing the possibilities of overstress failures under reliability marginal stresses, since they can assure lifetime reliability. But for new components like the lithium-ion battery, the failure mechanics as well as the stresses produced in the aircraft environment have been changed. Thus, we cannot presume what kind of failure mechanisms would occur due to chemical reaction until the projected lifetime reaches. Generally, chemical failure mechanisms are delicate and thus difficult to identify and reproduce, which means that the acceleration conditions and related factors can be hard to determine. Thus, they should test until lifetime under almost normal conditions, and the behavior of sufficiently degraded components should be checked under the rated stresses and overstresses. The media have reported various accidents or disasters due to unanticipated failure mechanisms in chemical items, such as the fires occurring in the Sony lithium-ion battery in notebook computers in 2005, Firestone tires causing Ford Explorer rollovers in the 1970s, wire bundles incorporating silver-plated copper wire leading to fire in the Apollo 1 cabin in 1967, and so on.

4 CONCEPTUAL FRAMEWORK OF SPECIFICATIONS FOR QUALITY ASSURANCE

Returning to the subject of establishing verification specifications, there are plenty of specifications that have few explanations technically. It is difficult to find articles that explain to establish specifications; there have been a few research studies about it. So let's consider how to anticipate issues in a new product and to configure a series of verification specifications responding to them, and how to develop specifications that will identify these issues accurately.

Here is one such methodology. First, select an existing product to be compared with the new one. All its relevant specifications are listed except the unnecessary specifications. Second, because inevitably the similar but older product has the intractable problems to be listed, we must devise new specifications to address these issues. Ongoing problems indicate that any counter-measures have not resolved the real cause because of the inadequate analysis. Nonetheless, the original design idea may be faulty. The existing product would be solved by using precise problem analysis, and the new product would be handled by identifying and fixing the problem before releasing the next model. To correct the existing problems in similar products, it is important to add the verification specifications. Sometimes the potential problems of the subassemblies manufactured by a new supplier also might be considered.

Third, the newly designed portions—those that differ from the current comparable product—should be listed and the potential issues related to them should be predicted. Verification specifications need to be devised to address these problems. Especially, note that all items incorporating new chemical materials should be tested to item lifetime under new quantitative specifications because a new kind of wear-out failure could occur near the item lifetime. Moreover, it is very difficult to computer-simulate and clarify diverse chemical reactions over an item's lifecycle.

Finally, the new product will also have performance fundamentals unique to it, which sometimes provide a competitive edge over competitors. Such comparative advantages in performance fundamentals might be checked with the newly established specifications.

A complete set of verification specifications in [Figure 11.7](#) might be summarized as four types of data: (1) all the verification specifications for the comparable product(s); (2) specifications to fix existing problems in the comparable product; (3) specifications that deal with the potential issues in the newly designed portions; and (4) specifications checking newly incorporated performance features. The specifications responding to the latter three categories are all established anew. The purpose of sectioning potential issues in a new product is to check whether necessary issues have been omitted or not.

All specifications enumerated according to this model would be classified, initially, into two groups: performance specifications and reliability specifications. If an issue to be identified is related to material rupture or degradation, especially over

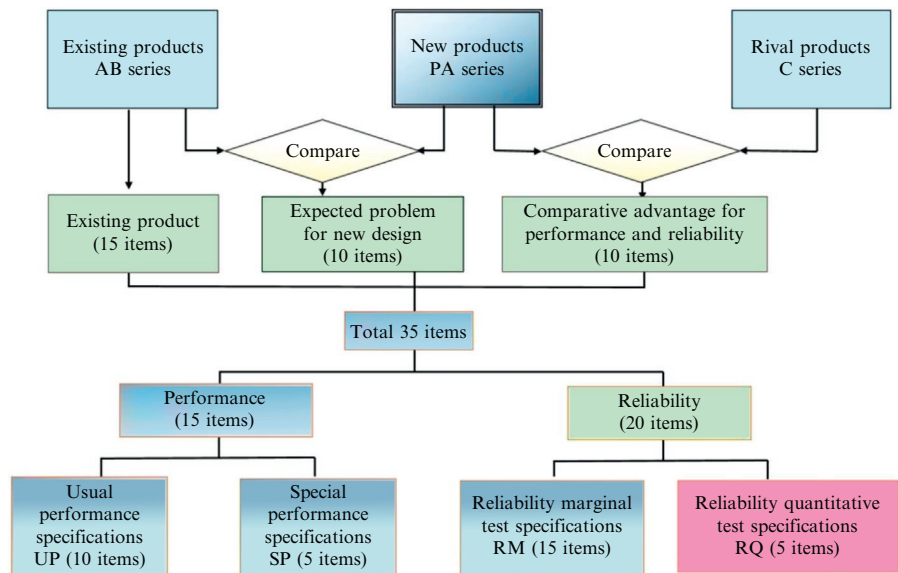


FIGURE 11.7

Complete testing sets of verification specifications for quality assurance.

time, it is a reliability issue; if not, it is a performance issue. The specifications are further divided into four categories: usual performance (UP) specifications, special performance (SP) specifications, reliability marginal (RM) test specifications, and reliability quantitative (RQ) test specifications.

UPs check the expected performance by the usual operator or consumer. SPs check performance under extraordinary environments, such as tropical heat or elevated electromagnetic fields. RMs are used for identifying physical changes under severe or peculiar conditions, including unusual usage environments like electrostatic overstress or lightning surges. Finally, RQs are for reviewing the product state under normal conditions and for estimating the product lifetime, the BX life (lifetime of the cumulative failure rate X%), and the annual failure rate within lifetime. Note that the lifetime index MTTF refers to the time-point at which about 60% of the production lot fails, which is an unacceptable rate.

5 CASE STUDY OF VERIFICATION SPECIFICATION: ROBOTICS

The purpose of this classification is to assign appropriate specialists to check whether requisite specifications have been evaluated sufficiently. Consider the format for Robot A in Table 11.1. Each line in the table represents one specification, which should be recorded. Specialists can browse the list, reviewing brief summaries for the adequacy of the verification specifications.

Table 11.1 Verification Specification Lists for Robot A

Category	Specification Name	Description	Sample Size	Criteria	Remark
UP1	Navigation test	Normal condition specified, moving distance accuracy	5	Max-Min < 10 mm	
SP1	High temperature/humidity test	60°/80% RH, 72 h	2	Works well	
RM1	Lightning test	10 kV-100 ms—3 times	2	Reset OK	
RQ1	Driving module life test	Current 4A (normal 2A)—40°, 1375 h	13	All works well	B2 life 5 years

UPs can be reviewed by development engineers, while SPs should be reviewed mainly by service and maintenance engineers; RMs are checked by reliability engineers who fully understand extraordinary states, and RQs should be reviewed by reliability engineers who especially understand parametric ALT.

Each specification must confirm the required functions under environmental and operational conditions, and list test conditions, test equipment with fixtures, test items, sample size, and criteria for test results. There are plenty of reference materials and specialists for configuring specifications regarding UP, SP, and RM. However, for RQ, there are few specialists and even fewer references to consult. Still, though it would seem almost impossible, it is not difficult to establish the specifications for RQ.

For an example, we will calculate a robot case. The robot incorporates one driving module using a direct current motor and gear train, for which the failure mechanism would be degradation of the motor wire and gear wear. Its lifetime is expected to continue over 5 years and its cumulative failure rate should be below 12% for up to 5 years. Assume that three-quarters of the failure rate (9%) is allocated to failures of other modules in the robot and that the remaining one-quarter (3%) pertains to the driving module, using the so-called part-count method. One-third of that (1%) is related to overstress failure of the drive module and the remaining failures (2%) are due to wear-out failure influencing the lifetime. Therefore, the lifetime target of the driving module will be expressed as B2 life 5 years—the cumulative failure rate, 2%, until 5 years. Assume that daily operation time is 3 h, which means 5500 h of operation in 5 years. The B2 life 5 years becomes B2 life 5500 h.

In order to reduce the testing time, the test load must be increased to flow the motor current by, say, a 100% more than usual, and the ambient temperature of the motor must also be elevated to, say, 10 °C higher than normal. This configuration accelerates the test speed by 8.0 times: current AF 4.0 power law and the temperature AF of 2. Thus, testing duration would be reduced to one-eighth or about 1 month (5500 h/8.0, 687.5 h).

The cumulative failure rate of 2% in the expression B2 life requires that we should test at least the inverse of 2% or 50 units. Two percent means two failures in 100 test items or one failure in 50 items. To attain the target B2 life 5500 h, we should test 50 units for 1 month under the above conditions. If no failure occurs, then the driving module meets the target with a common sense level of confidence. If we want to reduce the sample size, the test time must be increased. Doubling the test time to 2 months reduces the sample size one-quarter (square of the inverse of the test-time multiplier), or 13. The final test specification, then, is that 13 units should be tested for 2 months under elevated load and temperature conditions with the criterion that no failure is found. (To attain the overstress failure target of 1% per 5 years.)

With this simple procedure, it is possible for any engineer or stakeholder, even the CEO, to mentally calculate the sample size and testing time (based on target lifetime, AF, and test-time multiplier). All key personnel would be able to raise simple questions about test credibility, such as the validity of the sample size, test duration, AF, and so on, making corrections obvious and the new product more reliable. Afterward, it is enough for CEOs to check the metrics or reliability directly related to the outcomes of their businesses—the lifetime target and the annual failure rate.

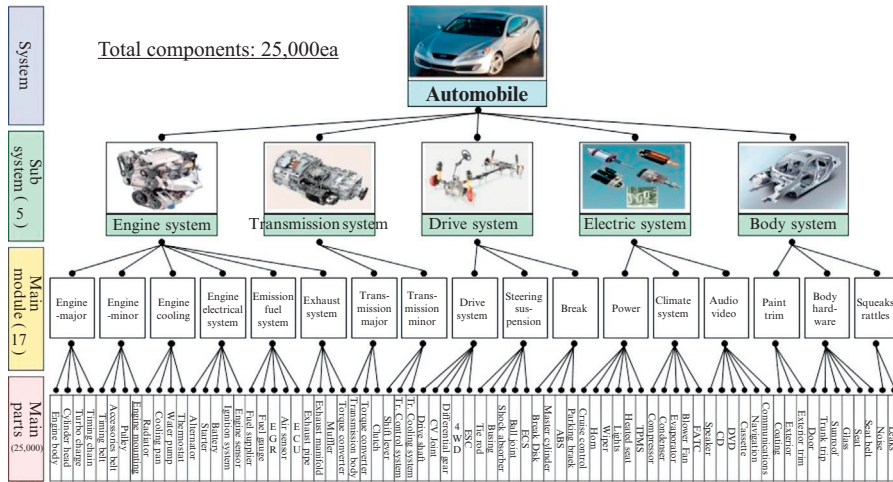
Note that with this method, we can obtain the reliability state of the production lot, but we do not know the time when and the site where each product will fail. This is because each item may have been made of a somewhat different material and by a slightly imperfect process. Furthermore, it is very difficult to estimate the next failure in item use because minute structural flaws caused by different materials and processes may have been magnified during use by various stresses that cannot be exactly identified. But it is not difficult to extend the lifetime of the production lot as an aggregate because we could test it longer, identify the failure site, and improve it with related correctives. Thus, it is easier to lengthen the lifetime of a certain item before its release into the market than to extend the lifetime of an item being used in the field.

6 PARAMETRIC ACCELERATED LIFE TESTING

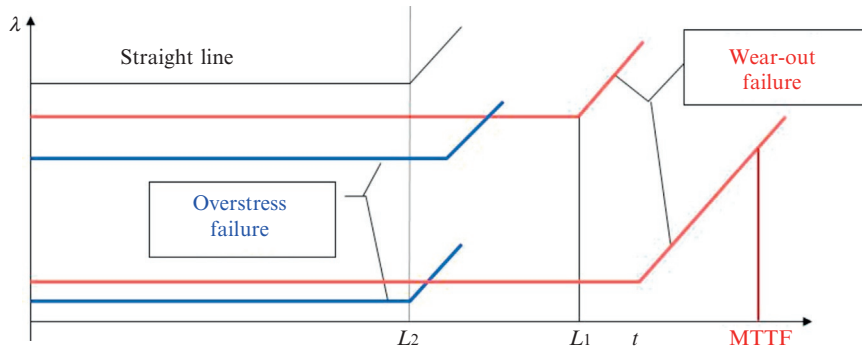
Parametric ALT for mechanical system like automobile in [Figure 11.8a](#) is an accelerated life testing to find the optimal design parameters not considering in design. First of all, it is critical to formulate the accelerated factor from the life-stress model and the dominant failure mechanism under the given physical and chemical conditions. Knowledge of physical of failure is required to understand the failure mechanism—traditionally fatigue or fracture in mechanical system due to repetitive stresses—related to a life-stress model.

Parametric ALT for RQ involves three key steps:

1. Create a life-stress model, determine the accelerating factor under severe conditions, and calculate the minimum testing periods required by dividing the target lifetime by the *AF*.



(a) Automobile field

(b) Product lifetime L_{BX} and failure rate λ_s **FIGURE 11.8**

Product lifetime such as automobile L_{BX} .

2. After assuming a shape parameter that is indicating the intensity of wear-out failure in the distribution of Weibull, calculate the necessary sample size considering the lifetime target.
3. Performing longer than minimum actual test periods, shape parameter will be determined.

Let us suppose that there is no initial failure, product lifetime L_2 due to overstress will determine in Figure 11.8b. The cumulative failure rate of set lifetime is the sum of the failure rate of each units.

6.1 ACCELERATED FACTOR

First of all, to create a life-stress model and accelerated factor, time-to-failure can be obtained from the McPherson's derivation [10],

$$TF = B[\sinh(aS)]^{-1} \exp\left(\frac{E_a}{kT}\right) \quad (11.2)$$

where T_f is time-to-failure, S is stress, E_a is activation energy, k is Boltzmann constant, T is absolute temperature, B and a are coefficients.

In medium stress, time-to-failure can be described as

$$TF = A(S)^{-n} \exp\left(\frac{E_a}{kT}\right) \quad (11.3)$$

The energy flow in a mechanical system can generally be expressed as efforts and flows (Table 11.2). Thus, it is judged that the stresses in mechanical or electrical system may come from the efforts (or loads) like [11].

For a mechanical system, when replacing stress with effort, the time-to-failure can be modified as

$$TF = A(S)^{-n} \exp\left(\frac{E_a}{kT}\right) = A(e)^{-n} \exp\left(\frac{E_a}{kT}\right) \quad (11.4)$$

The more severe the conditions, the shorter the testing time can be. This is a key factor of an accelerated life testing. In accelerated testing, the appropriate stress levels (S_1 or e_1) will fall outside the specification limits but inside the operating limits.

Under accelerated stress conditions, the AF can be described as:

$$AF = \left(\frac{S_1}{S_0}\right)^{-n} \left[\frac{E_a}{k} \left(\frac{1}{T_0} - \frac{1}{T_1}\right)\right] = \left(\frac{e_1}{e_0}\right)^{-n} \left[\frac{E_a}{k} \left(\frac{1}{T_0} - \frac{1}{T_1}\right)\right] \quad (11.5)$$

We can say that the first term is the outside effort (or load) and the latter is the internal energy in Equation 11.5. The equation has three unknown values—constant A and two parameters which are temperature and efforts. These parameters can be found using a three-level test under accelerated conditions. Under severe conditions, repetitive stress (or load) can be expressed as the duty effect that carries the on/off cycles and shortens unit lifetime [12]. Now, we derive the sample size equation to perform the parametric ALT.

Table 11.2 Effort and Flow in the Multi-Port System

Units	Effort, $e(t)$	Flow, $f(t)$
Mechanical translation	Force component, $F(t)$	Velocity component, $V(t)$
Mechanical rotation	Torque component, $\tau(t)$	Angular velocity component, $V(t)$
Compressor	Pressure difference, $\Delta P(t)$	Volume flow rate, $Q(t)$
Electric	Voltage, $V(t)$	Current, $i(t)$

6.2 SAMPLE SIZE EQUATION

The Cumulative Distribution Function (CDF) in Weibull can be expressed as:

$$F(t) = 1 - e^{-\left(\frac{t}{\eta}\right)^\beta} \quad (11.6)$$

Where t is time and η is characteristic life.

The Weibull reliability function, $R(t)$, is expressed as:

$$R(t) = e^{-\left(\frac{t}{\eta}\right)^\beta} \quad (11.7)$$

The characteristic life η_{MLE} from the Maximum Likelihood Estimate (MLE) can be derived as:

$$\eta_{MLE}^\beta = \sum_{i=1}^n \frac{t_i^\beta}{r} \quad (11.8)$$

Where n is sample size, t_i is testing duration for each sample, and r is the failure numbers.

If confidence level is $100(1 - \alpha)$ and the number of failure is $r \geq 1$, characteristic life, η_α , would be estimated from Equation 11.8,

$$\eta_\alpha^\beta = \frac{2r}{\chi_\alpha^2(2r+2)} \cdot \eta_{MLE}^\beta = \frac{2}{\chi_\alpha^2(2r+2)} \cdot \sum_{i=1}^n t_i^\beta \quad \text{for } r \geq 1 \quad (11.9)$$

Presuming there is no failures, p -value is α and $\ln(1/\alpha)$ is mathematically equivalent to Chi-Squared value, $\frac{\chi_\alpha^2(2)}{2}$. Characteristic life, η_α , would be represented as:

$$\eta_\alpha^\beta = \frac{2}{\chi_\alpha^2(2)} \cdot \sum_{i=1}^n t_i^\beta = \frac{1}{\ln \frac{1}{\alpha}} \cdot \sum_{i=1}^n t_i^\beta \quad \text{for } r = 0 \quad (11.10)$$

We know that Equation 11.9 will be established to all cases $r \geq 0$ and redefined as follows:

$$\eta_\alpha^\beta = \frac{2}{\chi_\alpha^2(2r+2)} \cdot \sum_{i=1}^n t_i^\beta \quad \text{for } r \geq 0 \quad (11.11)$$

To evaluate the Weibull reliability function in Equation 11.7, the characteristic life can be converted into B_X life as follows:

$$R(t) = e^{-\left(\frac{L_{BX}}{\eta}\right)^\beta} = 1 - x \quad (11.12)$$

After logarithmic transformation, Equation 11.12 can be expressed as:

$$L_{BX}^\beta = \left(\ln \frac{1}{1-x} \right) \cdot \eta^\beta \quad (11.13)$$

where L_{BX} is B_X life and x is cumulative failure rate till lifetime ($x = X/100$).

If the estimated characteristic life of p -value α , η_α , in Equation 11.11, is substituted into Equation 11.13, we obtain the B_X life equation:

$$L_{BX}^\beta = \frac{2}{\chi_\alpha^2(2r+2)} \cdot \left(\ln \frac{1}{1-x} \right) \cdot \sum_{i=1}^n t_i^\beta \quad (11.14)$$

Most lifetime testing has insufficient samples. The allowed number of failures would not have as much as that of the sample size.

$$\sum_{i=1}^n t_i^\beta = \sum_{i=1}^r t_i^\beta + (n-r)h^\beta \geq (n-r)h^\beta \quad (11.15)$$

After Equation 11.15 inserts into Equation 11.14, B_X life equation may be modified as follows:

$$L_{BX}^\beta \geq \frac{2}{\chi_\alpha^2(2r+2)} \cdot \left(\ln \frac{1}{1-x} \right) \cdot (n-r)h^\beta \geq L_{BX}^{*\beta} \quad (11.16)$$

Then, sample size equation with the number of failure will be modified as:

$$n \geq \frac{\chi_\alpha^2(2r+2)}{2} \cdot \frac{1}{\left(\ln \frac{1}{1-x} \right)} \cdot \left(\frac{L_{BX}^*}{h} \right)^\beta + r \quad (11.17)$$

For a 60% confidence level, the first term $\frac{\chi_\alpha^2(2r+2)}{2}$ in Equation 11.17 can be approximated to $(r+1)$ [13]. And if the cumulative failure rate, x , is below about 20%, the denominator of the second term $\ln \frac{1}{1-x}$ approximates to x by Taylor expansion. Then, the general sample size equation can be approximated as follows:

$$n \geq (r+1) \cdot \frac{1}{x} \cdot \left(\frac{L_{BX}^*}{h} \right)^\beta + r \quad (11.18)$$

If the accelerated factors in Equation 11.5 are added into the planned testing time, Equation 11.18 will be modified as:

$$n \geq (r+1) \cdot \frac{1}{x} \cdot \left(\frac{L_{BX}^*}{AF \cdot h_a^a} \right)^\beta + r \quad (11.19)$$

where h_a is actual testing time and AF is accelerated factor.

7 CONCLUSION

As those discussed above since 2000, about 800 specifications have been established, each of which has RQ as well as RM and performance test specifications (UP/SP), and have been applied to establishing these standards. Yet RQ test specifications would be not easy to be found in any specifications. And they can find the missing design parameters from the failure mechanism [14–21].

Product design consists of performance/reliability design itself and verifies their designs. For this, the verification groups as an independent party is separated from

the design groups. If the design people execute the verification on their own product, there is a very real possibility of failure, because their focus is on implementing good performance, not searching for problems. That is a matter for verification. Product verification might be performed by an independent division within the company or by an outside organization in order to attain success with the product and to widen its utility.

Furthermore, the tasks of verification are divided into two categories—establishing verification specifications and conducting verification activities. If a failure occurs, this makes it easy to trace whether mistakes lie in the specifications or in verification execution.

Verification specifications and their activities must be strictly performed as mentioned before. The deliverables by specification activities include specification name lists and specifications, both established and newly designed for the particular product. Test specification RQ of the lifetime under normal conditions using Parametric ALT might also be necessary. The verification specifications might be categorized into the four groups and reviewed by the appropriate specialists.

If the provision “Confirmation of Verification Specification” were inserted into NASA’s System Engineering Handbook, many failures and issues would be eliminated. It would also boost the development of the methodology for verification specification as well as quality improvement of products.

No matter how high the organizational capability is leveled up, there is no guarantee that a product’s quality has been improved to the highest level without individual product specifications. Only specification testing, including quantitative specifications fit to the new product, will accomplish it. Developing complete and accurate specifications will prevent future failures and guarantee a product’s value.

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A nonlocal damage-mechanics-based approach suitable for failure assessment and remaining life estimation of critical industrial components

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1 INTRODUCTION

For safety-critical plant components made of ductile materials, finite-element (FE) approaches which use material damage constitutive models are able to predict the failure behavior and the corresponding process with high accuracy [1–3]. For structural integrity assessment under various postulated and design loading conditions, the fracture resistance data of the base metal, weld metal as well as dissimilar welded joints, in terms of J - R curves are required not only in the upper-shelf, but also in the ductile-to-brittle transition (DBT) temperature regime [4–6]. Due to the accrual of irradiation damage in the nuclear reactor environment, the ductile materials may exhibit DBT behavior even at room temperature. It is well established that the fracture resistance behavior depends not only on the specimen geometry, but also upon the specimen size, crack-depth, loading and boundary conditions in the upper-shelf as well as in the DBT temperature regime [2,3].

Experiments may only be conducted on standard fracture mechanics specimens in the laboratory to determine the J - R curves as it will not only be expensive and time-consuming, but also be impossible in several situations to conduct fracture tests on actual reactor components with small-size surface cracks under actual loading conditions. Damage mechanics approaches [7,8] offer an impressive alternative to predict the J - R curves of the components with various postulated cracks and loading conditions. In the last three decades, damage-mechanics-based models like Rousselier's [7], Gurson's model (and its modified versions) [8,9] have become powerful tools in the safety analysis of nuclear reactor components, for example, pressure vessels, shell-nozzle junctions, pipings, headers, elbows, etc. These damage models predict the material behavior on the basis of the micromechanical processes (i.e., void nucleation, growth, and coalescence) leading to ductile fracture.

The conventional local damage-mechanics approaches suffer from the problem of mesh-dependency of the results [10–16] and it does not allow the use of small-size elements near the crack-tip (for a converged solution) to simulate large stress gradients in the DBTT regime. Nonlocal regularization of the material state variables can alleviate this problem and this has been investigated by various researchers over the years [15–19]. Recently, the authors have developed a nonlocal version of the Rousselier's damage model and showed that the results of this model are mesh-independent [17–19].

The ability of the nonlocal damage models to predict the effect of specimen size, geometry, crack-depth, loading and boundary conditions, etc., on the load-displacement and fracture resistance behavior has not been studied so much in detail in the research literature. In this chapter, we use the nonlocal Rousselier's model to

investigate these effects on the response of two different types of fracture mechanics specimens in the upper-shelf as well as the DBTT regime. The following issues and their solution in the context of use of nonlocal damage models have been discussed. The dependence of fracture resistance behavior and scatter of fracture toughness in the DBTT regime on the following parameters has been studied, that is,

- Crack depth (shallow crack vs. deep crack);
- Specimen geometry (compact tension “CT” vs. single-edged-notched-bend “SEB”);
- Specimen size (1T, 2T, and 4T CT specimens);
- Specimen thickness (keeping all other dimension unchanged)
- Boundary conditions (symmetric models vs. full model)

For simulation of probability of fracture in the DBTT regime, the combined nonlocal Rousselier’s damage model [17–19] and Beremin’s cleavage fracture model [20–22] was used. It was shown that local damage models are not able to predict several of these effects including the important aspect of the use of symmetric boundary conditions in FE analysis.

Almost all the specimens (especially, the fracture mechanics specimens) are symmetric to the expected crack plane in the mode-I loading case. In FE calculations, this symmetry is usually exploited and hence, only the half of the specimen is included in the geometrical model. In local damage-mechanics calculations, the damage tends to localize at the integration points. As a consequence, the predicted crack moves through the interior of the element (and not along the element borders as expected). This is in contradiction to experimental observation where the crack propagation path is on the element borders (and not through the element) in a symmetrically loaded specimen, unless there are heterogeneities in material property and micro-defect distribution in certain planes or directions. It was noted that there is a major difference between the results of local and nonlocal models when the symmetric boundary conditions are used.

In the DBTT regime, fracture toughness tests were conducted according to ASTM E-1921 standard [23] and the variation of reference temperature T_0 with respect to specimen size and geometry was studied. It was observed that the nonlocal damage models are able to satisfactorily predict the fracture resistance behavior as observed in the experiments and the effects of specimen size, geometry, crack-depth, loading and boundary conditions on the J - R curve in the upper-shelf and probability of fracture in the DBTT regime were also accurately predicted. It was also observed that the reference temperature T_0 does not have a fixed value as used in the master-curve approach [6]. The variation of T_0 for various types and sizes of specimens were predicted by the FE simulation and these results were compared with those of experiment. The advantage of nonlocal models is that the predicted results are not dependent upon the mesh size near the crack-tip. It is also able to accurately predict the effect of specimen geometry, size, loading and boundary conditions on the fracture resistance behavior.

Dissimilar metal welds impose a challenge to the engineers concerned with the structural integrity assessment of these joints. This is because of the highly

inhomogeneous nature of these joints in terms of their microstructure, mechanical, thermal, and fracture properties. Fracture mechanics based concepts cannot be directly used because of the problems associated with the definition of a suitable crack-tip loading parameter such as J -integral, CTOD, etc. Again, depending upon the location of initial crack (i.e., base, weld, buttering, different interfaces, etc.), further crack propagation can occur in any of these materials. Moreover, for interface cracks, there can be change in crack path from one material to the other.

The complex conditions and properties of the welded joint, as resulting from the elaborate interaction of different microstructures with gradients in material properties, have limited the ability of currently existing methods to construct the assessment on the basis of actual failure mechanisms of bimetallic welds. At present, the knowledge of homogeneous welded joints is utilized for design, and no specific methods exist for appraising approvable assessment of structural integrity. Simplified engineering treatment models have been developed in literature [24] for resolving the problem of dissimilar welds utilizing the concept of equivalent properties so that the integrity assessment methods for single materials can be used for them. However, it may be argued that the interaction between under-matched and over-matched local microstructures is one of the key elements prohibiting the optimized estimation of integrity of dissimilar metal welded components.

In this chapter, micromechanical models have been used for simulation of ductile fracture initiation and propagation of cracks in the dissimilar metal welds. The application of these damage models to predict fracture resistance behavior of dissimilar welded joints is not systematically investigated in research literature. The approach involves determination of micro-mechanical parameters of different material regions of the joint, prediction of fracture resistance behavior of the joint with initial crack located at different regions and validation by experiment.

The micro-mechanical parameters of four different materials, that is, ferrite, austenite, buttering, and weld have been determined individually by simulation of fracture resistance behavior of SEB specimens and comparing the simulated results with those of experiment. In order to demonstrate the effectiveness of the damage model in predicting the crack growth in the actual dissimilar metal welded specimen, simulation of a SEB specimen with initial crack at ferrite-buttering interface has been carried out.

For simulation of ductile and cleavage fracture, different types of yield functions and failure probability functions are used in literature [7–9,20]. A summary of different models which are used in the DBTT regime is provided in Table 12.1 and effects of different parameters on fracture behavior in upper shelf and DBTT regime are provided in Table 12.2.

2 NONLOCAL ROUSSELIER'S DAMAGE MODEL

The nonlocal formulation of Rousselier's damage model uses nonlocal damage d as a nodal degree of freedom in the FE mesh [17–19]. The increment of the nonlocal damage variable $\dot{d}$ in a material point $\vec{x}$ is mathematically defined as a weighted average of the increment of the local void volume fraction $\dot{f}$ in a domain Ω , that is,

Table 12.1 Summary of Different Models Used for Analysis of Ductile and Cleavage Fracture

Type of Model	Applicability (Ductile/Cleavage Fracture)	Yield Criteria/Failure Probability	For Details, Refer Section/References
Rousselier's damage model	Ductile fracture	$\phi = \frac{q}{1-d}$ $+ D\sigma_k d \exp\left(\frac{-p}{(1-d)\sigma_k}\right) - R(\epsilon_{eq}) = 0$	2, [7]
Gurson-Tvergaard-Needleman's model	Ductile fracture	$\phi = \left(\frac{q}{R(\epsilon_{eq})}\right)^2$ $+ 2q_1 f^* \cosh\left(-1.5q_2 \frac{p}{R(\epsilon_{eq})}\right) - 1 - q_3 f^{*2} = 0$	[8,9]
Beremin's model	Cleavage fracture	$P_f = 1 - \exp\left[-\left(\frac{\sigma_w}{\sigma_u}\right)^m\right],$ $\sigma_w = \sqrt[m]{\sum_{i=1}^n (\sigma_i)^m \frac{V_i}{V_0}}$	3, [20]

$$\dot{d}(\vec{x}) = \frac{1}{\Psi(\vec{x})} \int_{\Omega} \Psi(\vec{y}; \vec{x}) \dot{f}(\vec{y}) d\Omega(\vec{y}) \quad (12.1)$$

where $\vec{y}$ is the position vector of the infinitesimally small volume $d\Omega$ and $\Psi(\vec{y}; \vec{x})$ is the Gaussian weight function given by

$$\Psi(\vec{y}; \vec{x}) = \frac{1}{8\pi^{3/2}l^3} \exp\left(-\frac{|\vec{x} - \vec{y}|^2}{4l^2}\right) \quad (12.2)$$

The length parameter l in Equation 12.2 determines the size of the volume, which effectively contributes to the nonlocal quantity and is related to the scale of the microstructure. The above integral nonlocal kernel holds the property that the local continuum is retrieved if $l \rightarrow 0$. By expanding $f(\vec{y})$ in Taylor's series and substituting in Equation 12.1, one can obtain the damage diffusion equation as

$$\dot{d} - \dot{f} - C_{\text{length}} \nabla^2 \dot{d} = 0 \quad (12.3)$$

where C_{length} is the characteristic length parameter of the material [17–19]. The yield function ϕ of the Rousselier's model [7] is modified by substituting the nonlocal damage d in place of the local ductile void volume fraction f as [17] follows.

$$\phi = \frac{q}{1-d} + D\sigma_k d \exp\left(\frac{-p}{(1-d)\sigma_k}\right) - R(\epsilon_{eq}) = 0 \quad (12.4)$$

Table 12.2 Summary of Different Parameters Studied for Their Effect on Fracture Behavior in Upper Shelf and DBTT Regime

Parameter Studied	Effect on Fracture Behavior in Upper Shelf/DBTT Regime	Types of Specimens and Geometries Used in the Tests	Summary of Effects	For Details, Refer Section
FE mesh size (0.1, 0.2, and 0.4 mm)	Upper shelf region	IT SEB	<i>Local model</i> : faster load drop, inferior <i>J-R</i> curve with reduced mesh size <i>Nonlocal model</i> : results are independent of mesh size	6.1
Crack depth, <i>a/W</i> ratios of 0.13 and 0.522	Upper shelf region	IT SEB	Load carrying capacity and fracture resistance smaller for deep-cracked specimen compared to shallow-cracked specimen	6.2
Specimen geometry and crack-tip constraint	Upper shelf region	IT SEB, IT CT, 2T CT, 4T CT	Load carrying capacity for 1T SEB lower compared to 1T CT. With thickness of CT specimen, the load carrying capability increases, however, fracture resistance becomes inferior	6.3
Boundary condition	Upper shelf region	1T CT, full FE model vs. half model	<i>Local model</i> : Results are different for full and half FE model with symmetric boundary condition <i>Nonlocal model</i> : Same result for both full and half FE models	6.4
Crack depth, <i>a/W</i> ratios of 0.13 and 0.522	DBTT regime	IT SEB	Fracture toughness values as well as scatter are high for the shallow-cracked specimen in the DBTT regime due to low crack-tip constraint	6.5
Extent of stable crack growth before cleavage fracture at -100 , -60 and -20 °C	DBTT regime	1T CT	At -20 °C, the stable crack growth is higher. This value increases from -100 to -20 °C in the DBTT regime	6.6

Specimen geometry and crack-tip constraint	DBTT regime	1T SEB, 1T CT	Fracture toughness as well as scatter are almost similar for both types of specimens, however, fracture toughness values for 1T SEB are slightly higher in the DBTT regime	6.7
Specimen thickness 10, 25, and 50 mm	DBTT regime	SEB	Fracture toughness as well as scatter values become higher as the specimen thickness decreases	6.8
Specimen geometry	DBTT regime	1T, 2T, 4T	Fracture toughness values become lower as the specimen size increases, scatter is almost unaffected	6.9
Material mismatch at the crack-tip and ligament	Upper shelf region	Ferrite-buttering interface for dissimilar metal welded joint	Crack at the interface grows toward the buttering side due to inferior fracture toughness properties of buttering material compared to ferrite	6.10

where D and σ_k are the parameters of the Rousselier's model and are constants for a material, p is hydrostatic pressure, q is von-Mises equivalent stress, $R(\varepsilon_{eq})$ is material resistance (which is a function of von-Mises equivalent strain ε_{eq}). With loading, the void volume fraction evolves from the initial void volume fraction f_0 (volume fraction of eligible second-phase particles responsible for nucleation of voids upon plastic deformation) in the material. At a critical void volume fraction f_c , the voids coalesce with each other and at the final void volume fraction f_f , the material points loses its stress carrying capability. Hence, the above three void parameters (i.e., f_0 , f_c , and f_f) are also the material properties of the damage model. For solving the boundary value problem of the nonlocal damage continuum, one needs to solve the partial differential Equation 12.3 along with the mechanical equilibrium equation

$$\nabla \cdot \sigma + f_b = 0 \quad (12.5)$$

and the associated boundary conditions, where σ is the Cauchy stress tensor and f_b is the body force per unit volume. For the nonlocal damage degree of freedom, the additional (i.e., Neumann) boundary condition is used and is expressed as

$$\nabla d \cdot n|_{\Gamma_d} = 0 \quad (12.6)$$

where $n|_{\Gamma_f}$ is normal to the boundary Γ_f . By discretizing the weak forms of the governing differential Equations 12.3 and 12.5, we obtain the FE equations in matrix form as [17]

$$\begin{bmatrix} K_{uu} + K_{NL} & K_{ud} \\ K_{du} & K_{dd} \end{bmatrix} \begin{Bmatrix} \Delta \hat{u} \\ \Delta \hat{d} \end{Bmatrix} = \begin{Bmatrix} f_m^{ext} - f_m^{int} \\ -f_d^{int} \end{Bmatrix} \quad (12.7)$$

It may be noted that the stiffness terms K_{ud} , K_{du} , and K_{dd} in the element stiffness matrix are contributions of the nonlocal formulation. The matrix K_{uu} represents the conventional stiffness of the FE and it represents the mechanical stiffness, which corresponds to the relationship of the nodal mechanical forces and nodal displacement vectors.

Similarly, K_{ud} represents the matrix, which produces nodal damage vector corresponding to applied mechanical force vector. The matrix K_{du} produces nodal displacement vector corresponding to an applied damage force vector and K_{dd} produces nodal damage vector corresponding to applied damage force vector, respectively. f_m^{int} and f_m^{ext} are the internal and external mechanical force vectors, respectively, whereas f_d^{int} is the internal damage force vector. The above-described nonlocal formulation of the Rousselier's damage model was implemented in an in-house FE-based code and used for numerical simulation of fracture behavior of various types of specimens as discussed in the later sections. For simulation of probability of cleavage fracture in the DBT regime, the nonlocal damage model was combined with Beremin's model for cleavage fracture which is described briefly in the following section.

3 BEREMIN'S MODEL FOR CLEAVAGE FRACTURE

Beremin's model [20] for cleavage fracture is based on the weakest link concept where the probability of fracture can be represented in general as follows.

$$P_f = 1 - \exp\left(-\int_V g(\sigma_1^i) \frac{dV}{V_{\text{ref}}}\right) \quad (12.8)$$

where V is the volume of the plastically deformed zone (an essential condition for slip induced nucleation of cleavage micro-cracks) in the component, V_{ref} is the reference volume which is usually taken as 0.001 mm^3 . The function $g(\sigma_1^i)$ expresses the probability of failure of an infinitesimal volume i (having volume dV) according to the expression

$$dP_f = g(\sigma_1^i) \frac{dV}{V_{\text{ref}}} \quad (12.9)$$

where σ_1^i is the maximum principal stress acting at a material point i in the plastically deformed region having volume dV . According to Beremin's model [20] (which uses Weibull's statistics for distribution of defects responsible for triggering cleavage according to Griffith's theory), the form of $g(\sigma_1^i)$ is defined as

$$g(\sigma_1^i) = \left(\frac{\sigma_1^i}{\sigma_u}\right)^m \quad (12.10)$$

where m and σ_u are the Weibull's shape and size parameters, respectively. Using Equation 12.10 in Equation 12.8, one can obtain the probability of cleavage fracture P_f at any given loading as [20]

$$P_f = 1 - \exp\left[-\left(\frac{\sigma_w}{\sigma_u}\right)^m\right] \quad (12.11)$$

where the loading parameter is defined as the Weibull stress σ_w and is expressed as [20]

$$\sigma_w = \sqrt[m]{\sum_{i=1}^n (\sigma_1^i)^m \frac{V_i}{V_0}} \quad (12.12)$$

In this chapter, Equation 12.11 is used to calculate the probability of cleavage fracture of the CT specimen at different loading levels and at different temperatures, where the Weibull stress will be calculated using the stress field ahead of the growing crack using Equation 12.12. The nonlocal Rousselier's damage model has been used to calculate the crack-tip stress field. The effect of evolution of damage (ductile void volume fraction) on the stress field is taken care of by this model and hence, the effect of prior stable crack growth before initiation of cleavage fracture is modeled implicitly through this framework.

4 TYPES OF FRACTURE MECHANICS SPECIMENS ANALYZED IN UPPER SHELF AND DBTT REGIME

The nonlocal formulation of the Rousselier's damage model was used for simulation of fracture resistance behavior of various types of specimens in the upper-shelf as well as the DBTT regime. There are various advantages of the nonlocal formulation

over the local damage model and these will be discussed in this section in the context of simulation of fracture mechanics specimen with different geometry, size, and loading conditions.

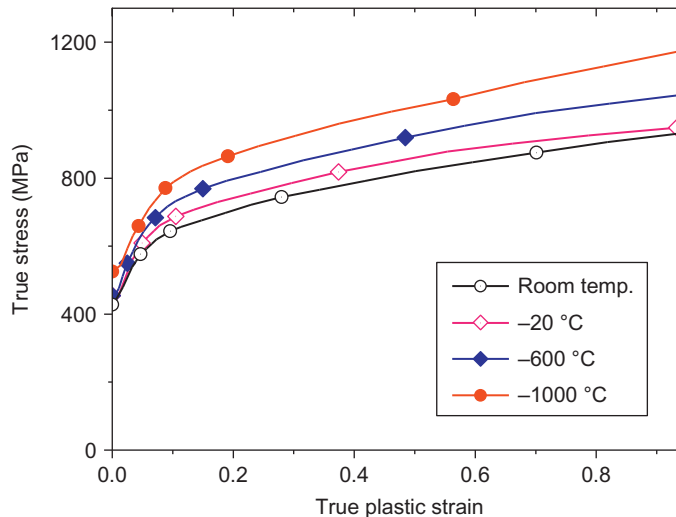
Initially, the standard 1T SEB specimen was simulated to obtain the load-displacement and the fracture resistance behavior at room temperature, where the fracture mode is purely of ductile nature. The effect of the FE mesh size (in the crack-tip region) on the load-displacement behavior and J - R curve of the 1T SEB specimen was studied using both local and nonlocal damage models. The objective is to demonstrate that the results of nonlocal formulation are independent of mesh size at the crack-tip and hence, this can be used further in simulation of fracture toughness variation and its scatter in the DBTT region, where both ductile and cleavage modes of fracture compete with each other before the specimen fails by unstable cleavage mode.

The effect of crack-depth of the specimen on the load-displacement response of the 1T SEB specimens was later studied by comparing the results of a shallow-cracked ($a/W=0.13$) specimen with that of a deeply cracked ($a/W=0.53$) specimen. Different sizes (i.e., 1T, 2T, and 4T) of SEB specimens were also analyzed. The effect of specimen geometry and loading condition on the load-deformation behavior was studied by comparing the results of simulation of the 1T SEB specimen with that of the 1T CT specimen. The effect of the use of symmetric boundary conditions on the results of local and nonlocal models was also investigated. All these studies were conducted initially at room temperature (upper-shelf region).

Subsequently, the scatter in fracture toughness and its variation with temperature was also simulated for various types of specimens in the DBTT region and the effects of the previously mentioned parameters (i.e., specimen geometry, size, crack-depth, loading conditions, etc.) were also studied. Experiments were conducted on all these different types of fracture mechanics specimens at room temperature as well as the DBTT regime and the results of FE analysis were compared with those of experiment. In order to demonstrate the effectiveness of the damage model in predicting the crack growth in the actual dissimilar metal welded specimen, simulation of a 1T SEB specimen with initial crack at the ferrite-buttering interface has been carried out.

5 MATERIAL AND EXPERIMENT

Experiments have been conducted on different types of fracture mechanics specimens for study of ductile crack growth in upper-shelf, lower-shelf as well as DBTT regime. The specimens have been machined from the pressure vessel steel DIN 22NiMoCr3-7. For FE analysis of the specimens, the material properties of the above material have been used. This material is homogeneous from the point of chemical composition and possesses good ductility. The true stress-strain curves of the material at different temperatures are shown in [Figure 12.1](#) and these are used in the FE analyses. Other material properties including the Rousselier's constants, characteristic length parameter, and the different void volume fractions (i.e., initial, final, and

**FIGURE 12.1**

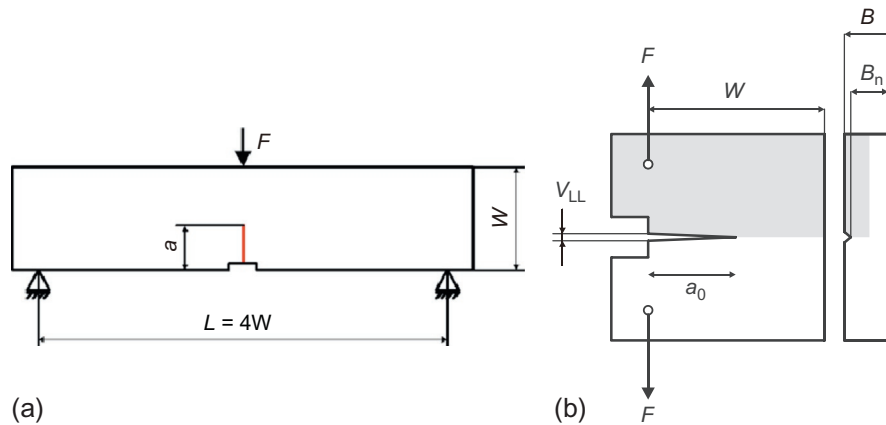
True stress versus true plastic strain curve of the material DIN 22NiMoCr3-7 at various temperatures.

at coalescence) are shown in Table 12.3. The 1T CT and SEB specimen has a thickness B of 25 mm, width W of 50 mm, and initial crack length “ a ” of 26.1 mm. The geometry, dimensions, loading and support conditions of the 1T SEB and 1T CT specimens are shown in Figure 12.2a and b, respectively. The specimens are 20% side-grooved and the a/W (initial crack length to width ratio) ratio is 0.52.

A dissimilar weld joint has been prepared between two pieces of ferritic and austenitic pipes (of 332 mm outer diameter and 29 mm thickness). The ferritic steel is DIN 20MnMoNi55 (similar to A508 Cl.3) and the austenitic steel is DIN 1.4550. The ferritic steel is an optimized reactor pressure vessel steel material. It has a bainitic structure at room temperature. From the metallographic examination of cross-sections of the deformed tensile specimens, it was observed that voids usually form

Table 12.3 Material Properties of DIN 22NiMoCr3-7 Pressure Vessel Steel

Model Parameters	Rousselier's Constants		Void Parameters			Char. Length Parameter C_{length} (mm ²)	E (Young's Mod. in GPa)	ν (Poisson's Ratio)
	D	σ_k (MPa)	Initial Void Volume Fraction f_0	Void Volume Fraction at Coalescence f_c	Final Void Volume Fraction f_f			
Value	2	445	0.0003	0.05	0.3	0.05	210	0.3

**FIGURE 12.2**

Geometry and loading conditions for ASTM standard specimens. (a) SEB specimen; (b) CT specimen.

from detachment of manganese sulfide (MnS) inclusions at the particle-matrix interface. The MnS particles inside the voids have almost a spherical shape. After nucleation of the voids from the MnS inclusions, there are no more adhesion between the particles and the matrix material. Therefore, the initial void volume fraction of the material can be equated to the volume fraction of MnS inclusions in the ferritic steel.

The austenitic material DIN 1.4550 (similar to SS304) is a piping material and has a face-centered cubic lattice crystal structure as opposed to the body-centered cubic lattice structure of the ferritic material. A metallographic cut through a deformed tensile specimen also exhibits voids in the cross-section. However, the voids usually develop at the niobium carbo-nitride inclusions, which normally have elongated shapes. The voids initiate mainly by the fracture of the slender carbo-nitride particles. It has also been observed that most particles are in already broken state as the piping material is usually subjected to plastic deformation during the manufacturing of the pipes. However, the adhesion between the particles and the matrix is much stronger than that of the ferritic material. The particles do not lie loose in the voids. Hence, the initial void volume fraction for the austenitic material does not correspond to the particle volume (but much lower) because the newly developed voids have only the same volume of the actual internally present cracks of the carbo-nitride particles.

After edge preparations on both the pipe ends, a buttering layer (of ~ 4 mm thickness) was deposited through a nickel-rich electrode on the ferritic pipe face. The buttering layer is of over-alloyed nickel-enriched E309L (24% Cr and 12% Ni). After this, welding is carried out between the buttering layer and the austenitic face of the pipe. The thickness of the weld is 25 mm approximately. The weld is deposited with matching E308L (18% Cr and 8% Ni) material. The use of nickel-enriched buttering layer is a common approach for power-plant applications to reduce toughness degradation as a consequence of carbon precipitation faced with matching austenitic filler materials.

Table 12.4 Chemical Composition of the Austenitic Steel (DIN 1.4550)

Element	C	Si	Mn	P	S	Cr	Ni	Mo	Nb	Fe
Wt. %	0.045	0.3	1.89	0.028	0.004	17.6	10.6	0.28	0.45	Rest

Table 12.5 Chemical Composition of the Ferritic Steel (DIN 20MnMoNo55)

Element	C	Si	Mn	P	S	Cr	Ni	Mo	V	Al	Cu	Fe
Wt. %	0.21	0.21	1.3	0.009	0.001	0.05	0.068	0.494	0.01	0.029	0.02	Rest

Table 12.6 Chemical Composition of the Weld Material

Element	C	Si	Mn	P	S	Cr	Ni	Mo	Nb	Ti	Fe
Wt. %	0.025	0.38	4.81	0.01	0.005	18.2	Rest	1.3	2.17	0.06	3.48

The regions critical for the performance of the weld are as follows, that is, the coarse-grained heat-affected zone, the fusion line, its immediate vicinity, and the buttering layer. Degradation of fusion zone toughness has been ascribed to the formation of upper-bainitic coarse-grained heat-affected zone microstructure, as well as a narrow martensitic layer (as a result of carbon diffusion from the ferritic steel toward the austenitic material). The chemical compositions of the austenitic, ferritic piping, and the weld material are given in [Tables 12.4–12.6](#), respectively. Other material properties such as Young's modulus of elasticity, yield stress, ultimate tensile strength (UTS), initial volume fraction f_0 , void fraction at saturated condition of nucleation f_n , percentage elongation, percentage reduction in area, Vickers hardness, etc. are shown in [Table 12.7](#). Various locations of the weld joint are etched to clearly locate the interfaces. An ammonium chloride mixture was used to etch the weld and the buttering regions. Electrolytic oxalic acid etching is also carried out for 2-3 min with 1.5-3.0 V electric potential with a cathodic area of 25 mm², respectively ([Table 12.6](#)).

Table 12.7 Parameters for Different Materials of the Dissimilar Weld Joint

Material	Young's Modulus (MPa)	Yield Stress (MPa)	UTS (MPa)	Initial Void Fraction f_0	Void Fraction at Saturated Nucleation Condition f_n	%age Elongation	%age Reduction in Area	Vickers Hardness
Austenitic steel	1.94E5	245	588	0.000001	0.0055	68.1	73.9	153
Ferritic steel	2.271E5	536.4	675	0.005	0.008	25.1	74.1	160
Weld material	1.82E5	448.1	587	0.001	0.008	14.2	17.9	234
Buttering material	2.0455E5	417.83	631	0.004	0	14.8	18.5	228

From the weld joint of the pipe, SEB specimens were machined. The sketch of the specimen with various dimensions is shown in Figure 12.2a. The proportion of dimensions of the specimen is in accordance to the ASTM E1820-99 standard. The pipe is welded in the circumferential direction. The specimens are machined such that the length of the specimen is in the longitudinal direction of the pipe, thickness of the specimen is in the thickness direction of the pipe, and the width of the specimen is in the circumferential direction of the pipe. As the outer diameter of the pipe is 332 mm, machining of a specimen of width 15 mm is easy. The specimen is a single-edge cracked and fatigue pre-cracked beam loaded in three-point bending and supported over a span S . The specimens were micro-etched in order to locate the metallurgically different microstructural zones and to machine initial cracks at the ferrite-buttering and the buttering-weld interface of the welded region of the specimen. The notch was initially prepared with saw-cutting and the finish cutting was done by electro-discharge machining. The initial crack size is of 15 mm. Later, fatigue pre-cracking done in the steps of 1 mm upto the crack size of 18 mm.

The specimens were 20% side-grooved in order to obtain a plane-strain condition for crack growth during subsequent loading. The crack mouth opening displacement (CMOD) was recorded through a crack opening displacement (COD) gauge which is attached to the data acquisition module of the universal testing machine, while the load was recorded directly from the load cells of the machine. The crack growth during the test was measured through the standard unloading (i.e., 20% unloading was done) compliance method. The J -integral for the specimens were calculated from the load-displacement data (i.e., using area under load-displacement curve and suitable η and γ factors) following standard ASTM method as described below.

The J -integral can be defined as the quantity which represents the rate of energy release with respect to the growing crack surface. According to Sumpter and Turner [25], the J -integral can be divided into an elastic and a plastic part, that is, J_{el} and J_{pl} whose expressions are written as

$$J = J_{el} + J_{pl} = \frac{K^2(1-\nu^2)}{E} + \frac{U_{pl}}{A_{net}} f\left(\frac{a}{W}\right) = \frac{K^2(1-\nu^2)}{E} + \eta \frac{U_{pl}}{A_{net}} \quad (12.13)$$

where K is the elastic stress intensity factor, ν is the Poison's ratio, E is Young's modulus of elasticity, A_{net} is the net cross-sectional area of the specimen that carries the load (i.e., product of effective thickness B_{net} considering side-grooving in the specimen and the remaining ligament length $(W - a)$, where W and a are width and remaining ligament, respectively), U_{pl} is the plastic part of the area of the load-displacement curve, η is geometric factor that depends upon the dimensions (i.e., initial crack length to the width ratio) of the specimen. As the elastic part of the area of the load-displacement curve U_{el} is very small compared to the plastic part of area, the J -integral can be evaluated in terms of the total area (i.e., $U_{tot} = U_{el} + U_{pl}$) and a correction factor (in terms of a geometric factor γ) is introduced to take into account of the increased crack length during crack growth. Hence, the expression of J -integral can be expressed as [26]

$$J = J_0 \cdot \left(1 - \frac{\gamma \cdot \Delta a}{(W - a_0)}\right) = \frac{\eta \cdot U_{tot}}{B_{net} \cdot (W - a_0)} \cdot \left(1 - \frac{\gamma \cdot \Delta a}{(W - a_0)}\right) \quad (12.14)$$

where a_0 is the initial crack length and Δa is the increment in crack length during loading (it increases as one traverses the load-displacement curve). The geometric factors η and γ depend upon the specimen geometry and crack length to width ratio (a/W). For a standard CT specimen, $\eta = 2 + 0.522(1 - a/W)$ and $\gamma = 0.75\eta - 1$. For a standard SEB specimen, $\eta = 2$ and $\gamma = 1$. The load-CMOD and the J -integral versus crack growth data for these specimens were obtained from the experiments and compared with results of numerical simulation.

6 RESULTS AND DISCUSSION

Results of FE analysis of various types of specimens as discussed in the previous section are presented in the following sections and the data are compared with those of experiment wherever applicable. A summary of different parameters studied in this work is given in Table 12.2.

6.1 EFFECT OF MESH SIZE ON THE LOAD DEFORMATION AND FRACTURE RESISTANCE BEHAVIOR OF THE 1T SEB SPECIMEN

In order to demonstrate that the results of the nonlocal damage model are mesh-independent, a standard 1T (thickness of 1 inch approximately) SEB specimen was analyzed. In the FE mesh, 8-noded iso-parametric elements were used and 2D plane-strain analysis was carried out. In the region of the crack-tip, quadratic square-shaped elements with different edge lengths (i.e., 0.1, 0.2, and 0.4 mm, respectively) were used. Figure 12.3 shows the load-CMOD response of the 1T

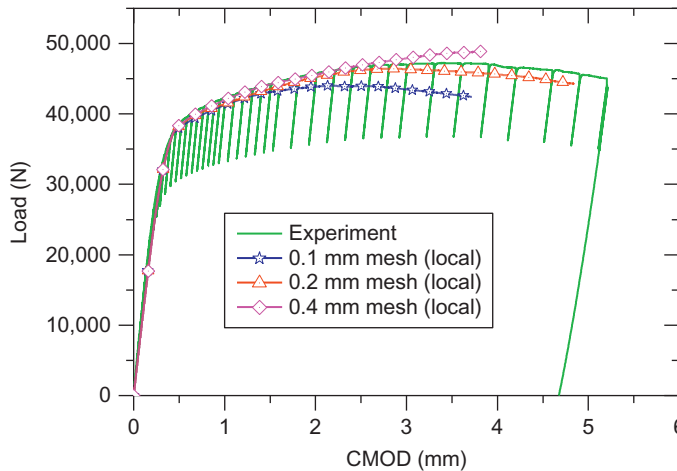


FIGURE 12.3

Effect of mesh size at the crack-tip on load-deformation behavior of a 1T SEB specimen (results of local damage model).

SEB specimen as computed using the local Rousselier's model. The CMOD is measured at the extreme end of the crack mouth. The results of analysis with different mesh sizes are also compared with the experimental data. It can be observed that results of the local model with mesh size of 0.2 mm compares very well with that of experiment though load values are slightly lower for the FE analysis for a given value of CMOD. This is because of the use of 2D plane-strain elements in the FE analysis where actual experimental condition corresponds to that of 3D.

The load-CMOD response for the case of 0.1-mm mesh size shows considerable softening behavior and the load falls much more rapidly compared to that of experiment. This is because of prediction of faster crack growth in the FE analysis compared to that observed during experiment. This is the characteristic of local damage or softening models where localization of damage takes place (due to loss of ellipticity of the governing differential equation) in a narrow region which is decided by the size of elements used in the analysis. The finer the mesh, the lesser the energy it takes for damage accumulation and hence, faster the crack growth. For the analysis with 0.4-mm element size at the crack-tip, the local damage model overestimates the load value for a given value of CMOD as the predicted extent of crack growth is lower compared to that actually observed in experiment. Hence, the results of local damage models are highly sensitive to the size of discretization in the crack-tip region. The use of nonlocal formulation with nonlocalization of the damage parameter avoids this problem as will be seen in the following sections.

The nonlocal Rousselier's damage model was used for the simulation of the load-CMOD response of the 1T SEB specimen with different mesh sizes and the results are shown in Figure 12.4. It can be observed that the load-CMOD response for all the mesh sizes are almost same and they also compare very well with those of

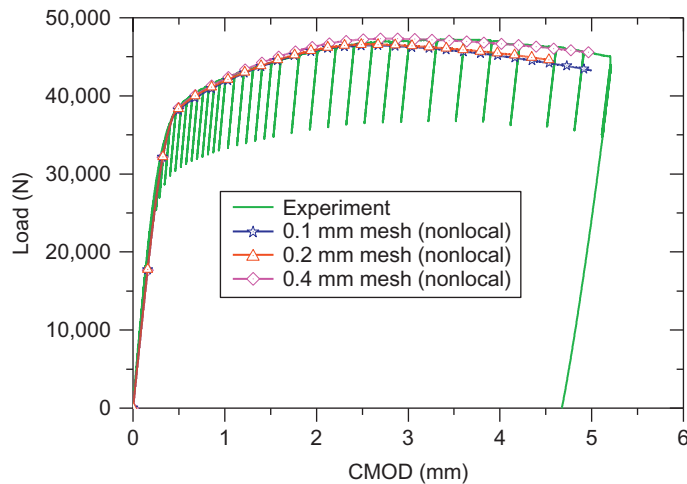
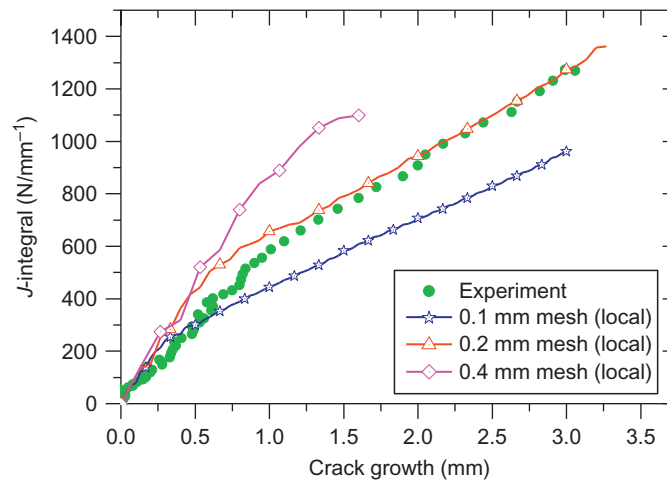


FIGURE 12.4

Effect of mesh size at the crack-tip on load-deformation behavior of a 1T SEB specimen (results of nonlocal damage model).

**FIGURE 12.5**

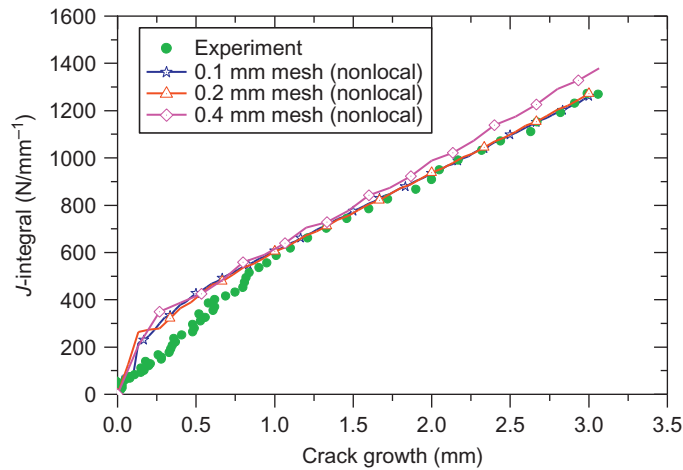
Effect of mesh size at the crack-tip on fracture resistance (J - R curve) behavior of a 1T SEB specimen (results of local damage model).

experiment. This is because the crack growth is very well predicted as a function of applied displacement. This aspect will be clear when the fracture resistance (J - R or J vs. crack growth) curve is evaluated from the area under the load-displacement curve of the 1T SEB specimen with ASTM standard techniques.

The J - R curves obtained for different mesh sizes with the use of local damage models are shown in Figure 12.5, whereas those obtained with the use of nonlocal damage models are shown in Figure 12.6 along with the experimental data. It may be observed that the results of the nonlocal model are mesh-independent and the FE results compare very well with the experimental data. The local model predicts different J - R curves for different mesh designs (i.e., mesh sizes) at the crack-tip and hence, it is not suitable for the prediction of fracture resistance behavior when one needs to use very fine mesh at the crack-tip in order to obtain a converged solution for the stress field. On the other hand, the nonlocal model doesn't suffer from this handicap and hence, it has later been used to predict the effect of specimen geometry, size, crack-depth, etc., on the load-deformation response in the upper-shelf as well as the DBTT regime and the results were also compared with those of experiment in order to verify the ability of the model to simulate the correct experimental response.

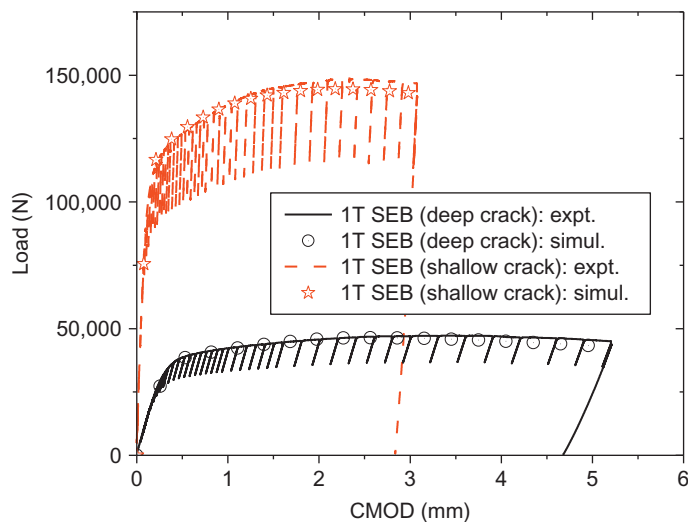
6.2 EFFECT OF CRACK DEPTH ON LOAD-DEFORMATION RESPONSE OF 1T SEB SPECIMENS

In order to study the effect of crack-depth on load-deformation response, experiments have been conducted on 1T SEB specimens with initial a/W ratios of 0.13 and 0.522, respectively. The standard 1T SEB specimen has a thickness B of 25 mm thickness, width W of 50 mm. The dimensions and loading conditions of the specimen are shown

**FIGURE 12.6**

Effect of mesh size at the crack-tip on fracture resistance (J - R curve) behavior of a 1T SEB specimen (results of nonlocal damage model).

in Figure 12.2a. The specimen is 20% side-grooved and hence, the net thickness in the uncracked remaining ligament is 20 mm. A 2D plane-strain analysis with 8-noded isoparametric quadrilateral elements has also been carried out. Due to symmetry, only half of the specimen has been analyzed. The load-CMOD response of the shallow-cracked and deeply cracked specimens are plotted in Figure 12.7 along with

**FIGURE 12.7**

Load-CMOD response of a 1T SEB deep-cracked versus shallow-cracked specimen (experiment and FE simulation).

corresponding experimental data. Same parameters, as used for the analysis of the 1T SEB deeply cracked specimen ($a/W=0.522$), have been used here for simulation of 1T SEB shallow-cracked specimen ($a/W=0.13$). A very good correspondence of the FE simulation results with those of experiment can be observed in Figure 12.7. Hence, the capability of the nonlocal model to accurately predict the effect of crack-depth on the load-CMOD response could be demonstrated in this work.

6.3 EFFECT OF GEOMETRY, SPECIMEN SIZE, AND LOADING CONDITION ON THE LOAD-DEFORMATION RESPONSE

It may be noted that 1T CT and 1T SEB specimen have different geometry and loading conditions. The crack-tip region of the CT specimen is predominantly loaded in tension (though some bending component is also present due to offset of the loading axis with respect to the crack-tip), whereas the crack-tip of the SEB specimen is mainly loaded in bending. We want to demonstrate that the parameters of the nonlocal model could not only predict the behavior of the 1T SEB specimen, but also could predict the response of the 1T CT specimen, where the crack-tip is loaded in a different state of stress. Figure 12.8 shows the simulated load-CMOD response of the 1T CT as well as the 1T SEB specimens as obtained from FE analysis. The results are also compared with those of experiment and one can observe a very good agreement between the results of FE analysis and those of experiment.

Similarly, CT specimens with different sizes were tested. These are denoted as 1T, 2T, and 4T, respectively. The dimensions of the 1T CT specimen were given earlier. The dimensions of the 2T and CT specimens are two and four times of the corresponding

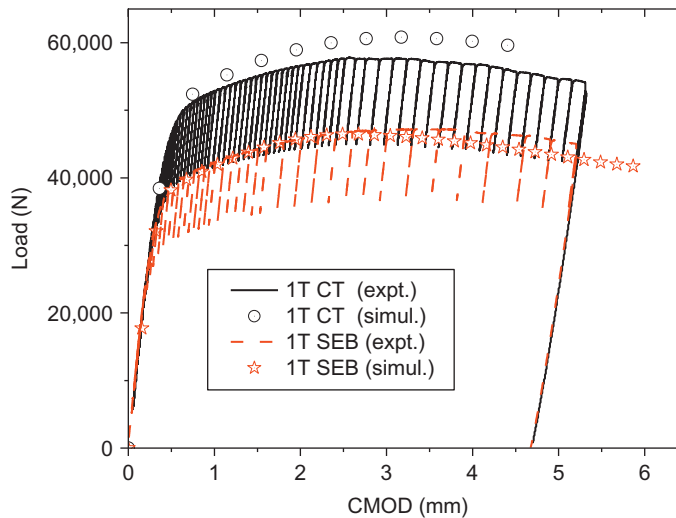
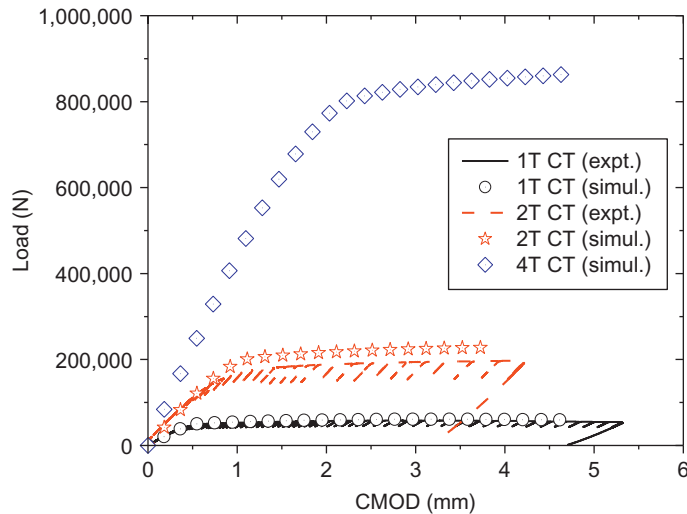


FIGURE 12.8

Load-CMOD response of a 1T CT versus 1T SEB specimen (experiment and FE simulation).

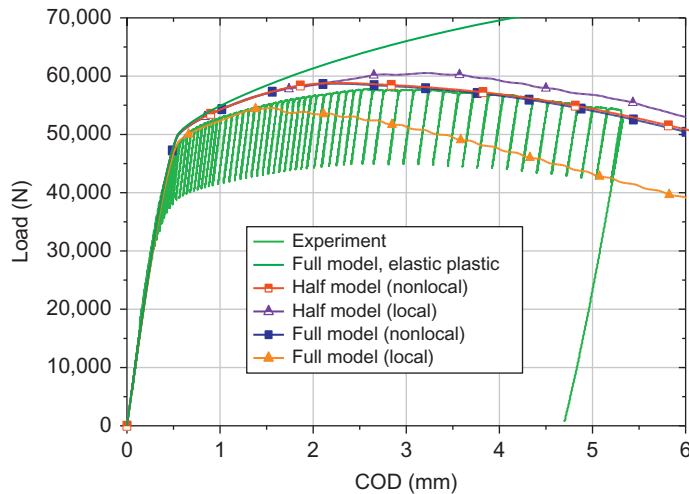
**FIGURE 12.9**

Load-CMOD response of a 1T, 2T, and 4T CT specimen (experiment and FE simulation).

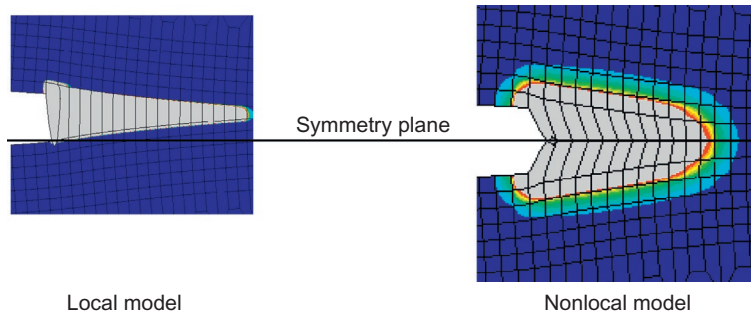
dimensions of the 1T CT specimens, that is, the specimens were just scaled by a factor of two and four, respectively. The aim of this exercise is to demonstrate that the same set of micro-mechanical parameters should be able to satisfactorily predict the load-CMOD responses of all these different sizes of CT specimens. The results of FE analysis with the nonlocal model are plotted in Figure 12.9 along with the experimental data and a very good agreement between the two sets can be observed.

6.4 EFFECT OF SYMMETRY BOUNDARY CONDITIONS ON CRACK GROWTH SIMULATION

Often, laboratory specimens and noncomplex components are built up and loaded symmetrically. For example, in the case of a CT specimen, it is sufficient to model and simulate half of the specimen if the symmetric nature of the boundary conditions is exploited. If the full model is analyzed instead of the half or the symmetric model, the results of analysis should not change. However, it was observed that the results of the local version of the Rousselier's damage model are not only mesh-dependent, but also dependent upon the choice of the boundary condition. Figure 12.10 shows the load-CMOD response of the 1T CT specimen as obtained from FE analysis when local model was used for the half and the full model, respectively. It can be observed that the two results are different. However, the nonlocal model predicts exactly the same result for both the cases. It is because of the localization of crack growth in one side of the symmetry line in case of the simulation with local model as can be seen from Figure 12.11. In case of FE simulation with the nonlocal damage model, material damage occurs in a symmetric fashion on both sides of the line of symmetry demonstrating that the crack grows along the center line when all the conditions for

**FIGURE 12.10**

Effect of symmetric boundary conditions on load-deformation behavior of a 1T CT specimen (half vs. full model used in local and nonlocal FE analysis).

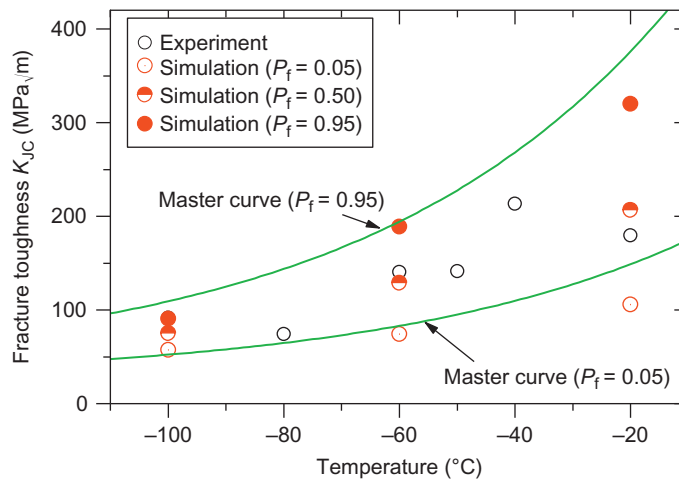
**FIGURE 12.11**

Crack growth path for a full 1T CT specimen (without symmetric boundary conditions, local vs. nonlocal model).

symmetry are met. Hence, the results of the nonlocal model are more realistic when compared to those of local model. Due to the above observation, one must be careful about the use of local damage models for prediction of crack path when the geometry, initial crack configurations, and loading conditions are complex.

6.5 EFFECT OF CRACK DEPTH ON THE PROBABILITY OF CLEAVAGE FRACTURE IN THE DBTT REGIME

In the DBTT regime, the crack-tip experiences a large stress gradient and hence, the ductile crack growth (of the order of few microns) before unstable cleavage fracture cannot be predicted by the local damage models. This is because of the mesh-dependent nature of

**FIGURE 12.12**

Predicted fracture toughness scatter in the DBTT region for the deep-cracked 1T SEB specimen.

the results obtained with these models. In order to use fine mesh in the crack-tip and at the same time, to use ductile damage models, the nonlocal formulation of the Rousselier's damage model has been used in this work. The results of this model have been demonstrated to be mesh-independent in the previous sections. Hence, this model along with the Beremin's model for cleavage fracture has been used to simulate the fracture transition curve for two different types of materials.

The true stress-strain curve of the material at different temperatures as used in FE analysis is shown in Figure 12.1. Other material properties including the Rousselier's constants, characteristic length parameter, and the void volume fractions (initial, final, and at coalescence) are shown in Table 12.1. Experiments have been conducted on standard deeply cracked 1T SEB specimens at various temperatures in the DBTT regime of the material. The experimental data of fracture toughness K_{JC} has been plotted at various temperatures in Figure 12.12. It can be observed that there is a substantial scatter in the fracture toughness data.

The scatter is lowest in the lower-shelf temperature at -100 °C and it increases with temperature. The scatter near the upper-shelf region is highest. This is because of the presence of two different processes (i.e., cleavage and ductile fracture) toward the upper-shelf temperature region in the DBTT regime. Each of the fracture processes has inherent scatter. The scatter in initiation of cleavage fracture is due to the scatter in size and distribution of second-phase particles which are responsible for triggering of the same (mainly carbides in these ferritic steels). In case of ductile fracture, the size, distribution, and interfacial strength of the inclusions (MnS type) and other second-phase particles control the scatter in fracture toughness. As both these processes operate near the right end of the DBTT regime, the scatter is larger compared to pure cleavage fracture in the lower-shelf region.

Figure 12.12 shows the predicted fracture toughness variation (with temperature) for the deeply cracked 1T SEB specimen. FE analysis has been carried out using 3D iso-parametric 20-noded brick elements. The mesh size near the crack-tip has been of the order of 0.02 mm. The Weibull parameters (m and σ_u) used in the analysis are 34.6 and 2005 MPa, respectively. The crack-tip stress field as computed by the non-local damage model has considered the effect of the ductile crack growth and hence, the predicted fracture toughness scatter compares very well with that of experimental data (Figure 12.12).

The master curves for fracture probability of 5% and 95% have also been plotted along with the experimental and FE results in Figure 12.12. It can be observed that the combined nonlocal ductile and cleavage fracture model has been able to predict the experimental data of fracture toughness as well as the master curve very accurately. However, the master curve requires the evaluation of the transition temperature T_0 from experimental data and hence, this T_0 may be not applicable when the fracture toughness variation of different types of specimens and components are to be predicted. It will be shown in the following sections that the numerical model as discussed in this work is able to predict the effect of specimen geometry, size, crack-depth, loading types, etc., on the shape of the fracture toughness transition in the DBTT region of ferritic pressure vessel steels.

Experiments have been conducted on both shallow-cracked ($a/W=0.13$) and deeply cracked ($a/W=0.53$) 1T SEB specimens. The fracture toughness variation in the DBTT regime has been calculated by the model for the shallow-cracked SEB specimen and the results are plotted in Figure 12.13 along with experimental data and the master-curve data. It can be observed that the FE model has been able to predict the experimental scatter very well in the whole DBTT region. The scatter in fracture toughness as predicted by the model for the deeply cracked specimen has

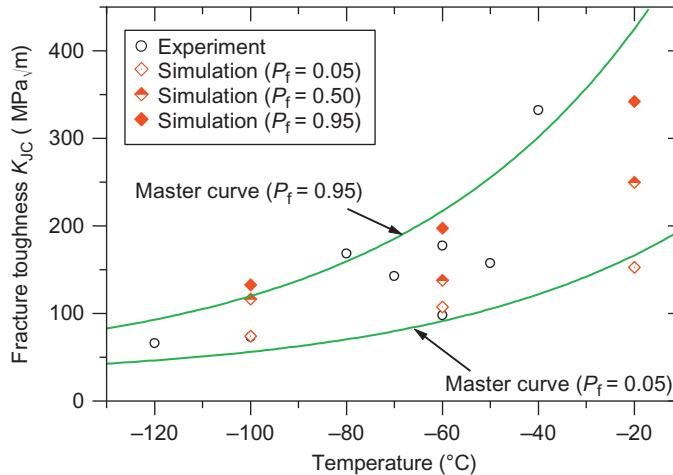
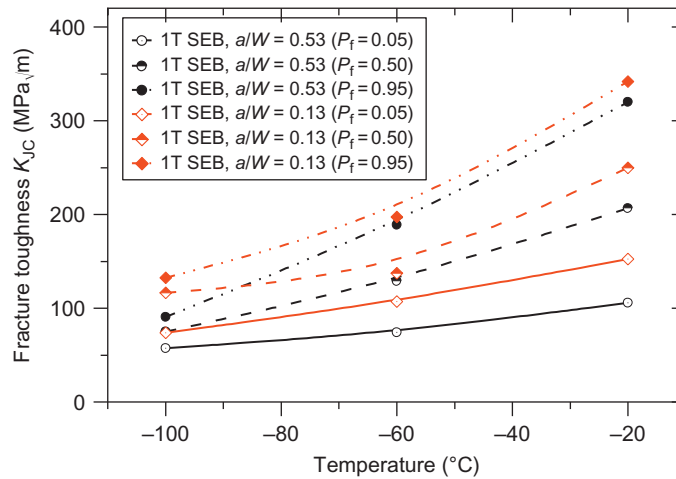


FIGURE 12.13

Predicted fracture toughness scatter in the DBTT region for the shallow-cracked 1T SEB specimen.

**FIGURE 12.14**

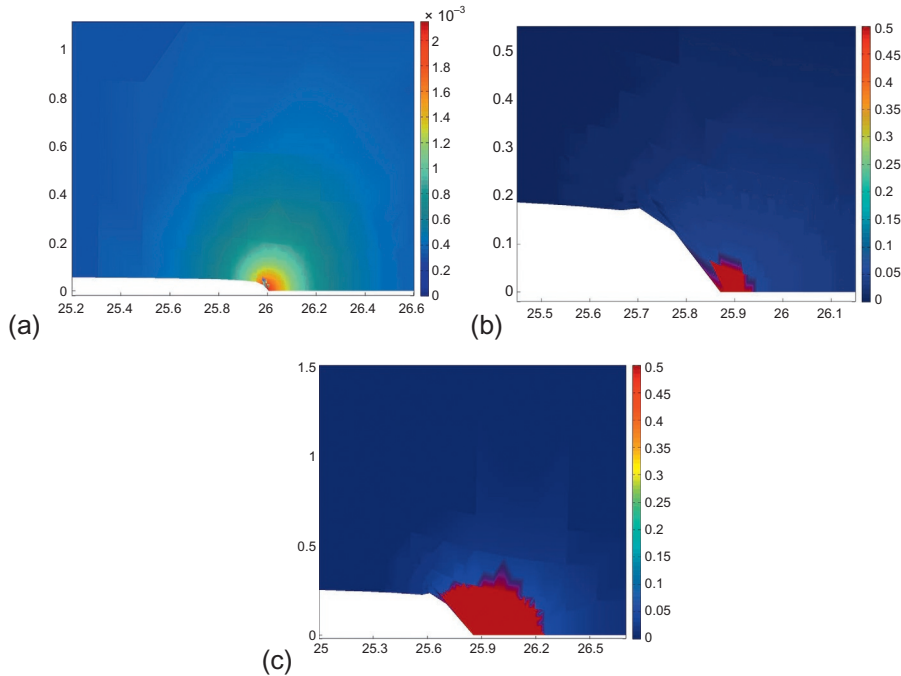
Effect of crack-depth on fracture toughness variation in DBTT regime (shallow vs. deeply cracked 1T SEB specimen).

been compared with that of the shallow-cracked specimen in Figure 12.14. The fracture toughness values of the shallow-cracked SEB specimen are considerably higher compared to the deeply cracked SEB specimen and hence, the transition temperature is also lower for the shallow-cracked specimen. For this reason, shallow-cracked specimens are less susceptible to DBT compared to the deeply cracked specimens. This is in line with experimental observation.

As the crack-tip constraint is lower for the shallow-cracked specimens, the magnitude of crack-tip stress field is lower compared to that of the deeply cracked specimens for the same level of loading. This results in lower values of Weibull stresses and hence, the probability of fracture is lower for a given value of K_{JC} loading in case of the shallow-cracked specimen. In components, the observed cracks are usually surface cracks with very small depth compared to the thickness and hence, the data of shallow-cracked specimens are more appropriate for evaluation of transition temperature in these cases. The master curve as evaluated from deeply cracked specimens cannot be applied directly for safety evaluation of actual components with very small surface cracks and hence, the modeling technique as presented in this work will be very useful for such situations.

6.6 EXTENT OF STABLE CRACK GROWTH (BEFORE CLEAVAGE FRACTURE) IN THE DBTT REGIME

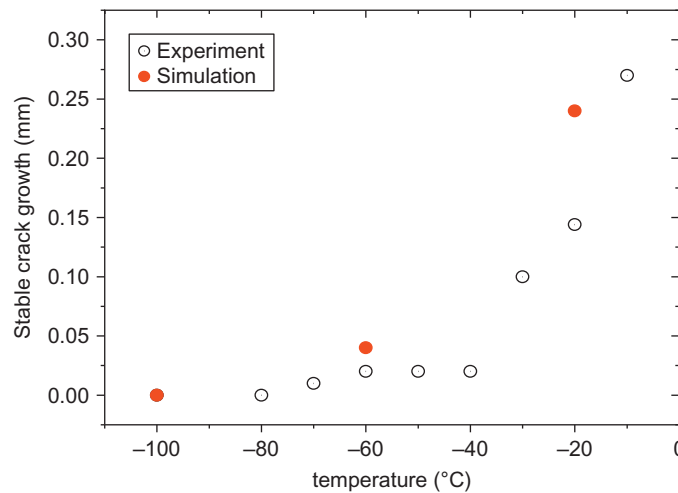
The extent of stable crack growth before cleavage fracture varies drastically with temperature in the DBTT regime. In the lower-shelf regime (-100°C), the fracture mode is purely cleavage and hence, the extent of stable ductile crack growth is

**FIGURE 12.15**

Extent of region with complete damage (damage value of 0.5) for 1T CT specimen at (a) -100°C , (b) -60°C , (c) -20°C .

absent. This is the reason why the scatter in the fracture toughness value is also less in the lower-shelf fracture regime. In order to demonstrate this behavior, stable crack growth in the standard 1T CT specimen in the DBTT region before the onset of cleavage fracture was computed using the nonlocal damage model. Figure 12.15 presents the contours of damaged regions ahead of crack-tip at three different temperatures (i.e., -100 , -60 , and -20°C , respectively).

The regions of stable crack growth are represented by regions of damage value exceeding 0.4 in the contour plots. It can be seen from Figure 12.15 that the extent of stable crack growth at -20°C is of the order of 0.2 mm, whereas the same is of the order of 0.05 mm at -60°C . At the lower-shelf, the maximum damage value is very small and hence, there are no regions of stable crack growth. This is consistent with experimental observation (the specimens fracture surfaces were studied after the tests and the extents of stable crack growth were measured). The experimentally measured values (average values) of stable crack growth (before unstable fracture) are compared with those of FE simulation in Figure 12.16 and it can be observed that the nonlocal damage model is able to predict the extent of stable crack growth very accurately and hence, the resulting fracture toughness scatter as predicted by the model in the DBTT regime matches closely with that of experiment.

**FIGURE 12.16**

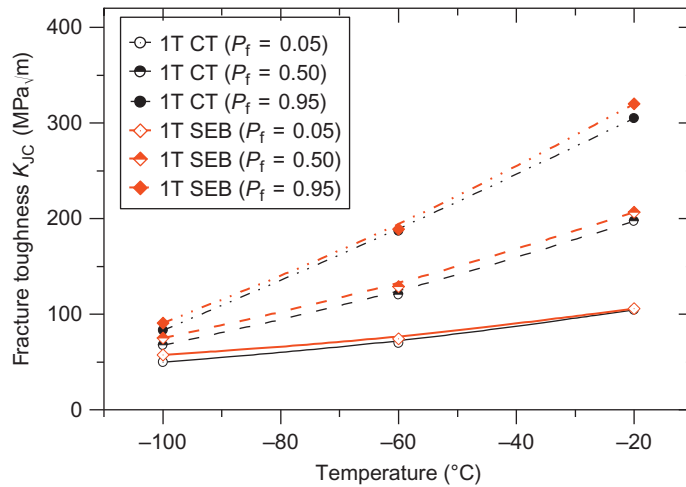
Variation of stable crack growth with temperature for a 1T CT specimen (exp. vs. FE simulation).

6.7 EFFECT OF SPECIMEN GEOMETRY AND LOADING CONDITION ON PROBABILITY OF CLEAVAGE FRACTURE IN THE DBTT REGIME

The fracture toughness transition curves as predicted by the model for both the 1T CT and 1T SEB specimens are plotted together in Figure 12.17. It can be observed that the 50% and 95% failure probability curves differ slightly among the two types of specimens. The two types of specimens see different crack-tip loading conditions and hence, the scatter in fracture toughness curves is slightly different. The scatter is more for the SEB specimen especially in the right-hand side of the DBTT regime. This indicates that the transition temperature for the SEB specimen is slightly lower for the 1T SEB specimen in comparison to the 1T CT specimen. This has also been observed in experiments where the transition temperature T_0 is evaluated according to ASTM E1921 standard. Hence, the prediction by the model is in line with experimental observation.

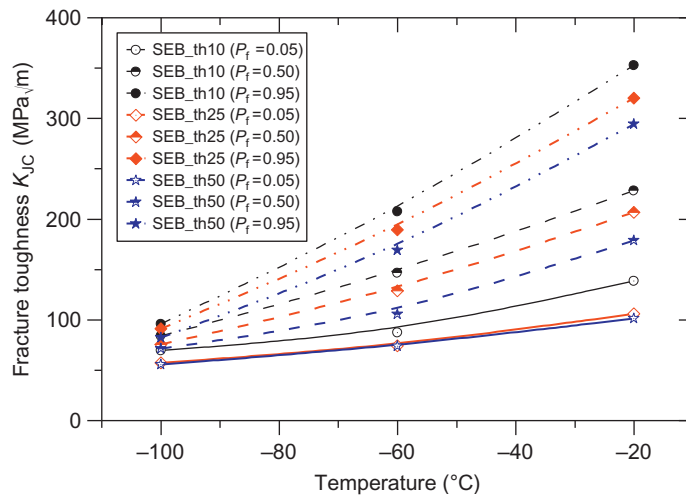
6.8 EFFECT OF SPECIMEN THICKNESS ON PROBABILITY OF CLEAVAGE FRACTURE IN THE DBTT REGIME

The effect of specimen thickness on the fracture toughness variation in the DBTT region was also studied by testing 1T SEB specimens. Specimens with thickness values of 10, 25, and 50 mm (keeping all other dimensions unchanged) were tested and the simulations were carried out with the help of the nonlocal damage model. Figure 12.18 presents the predicted results of fracture toughness variation (with temperature and probability of fracture) for specimens with three different values of thickness. It can be observed from Figure 12.18 that the scatter in the fracture

**FIGURE 12.17**

Effect of specimen geometry and loading condition on fracture toughness variation in DBTT regime (1T CT vs. 1T SEB specimen).

toughness decreases for the specimens with larger thickness values, signifying the increase in the fracture toughness transition temperature. Hence, specimens with larger thickness values are more susceptible to the DBT for a given value of loading. This is in line with the experimental observations.

**FIGURE 12.18**

Effect of specimen thickness (10, 25, and 50 mm) on fracture toughness variation in DBTT regime of a 1T SEB specimen.

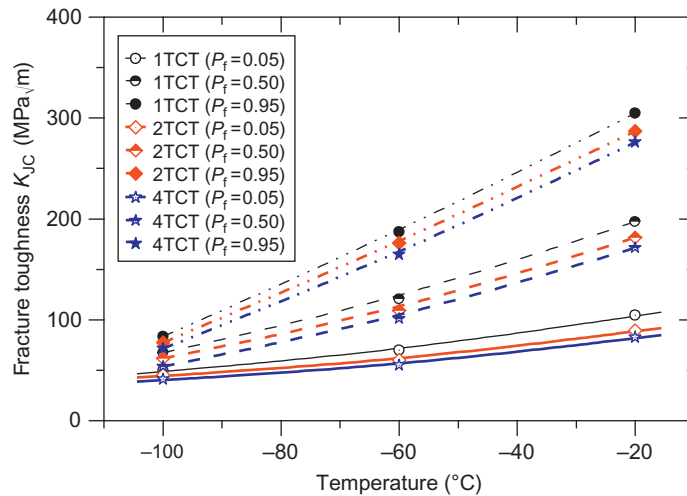


FIGURE 12.19

Effect of specimen size (1T, 2T, and 4T) on fracture toughness variation in DBTT regime of CT specimen.

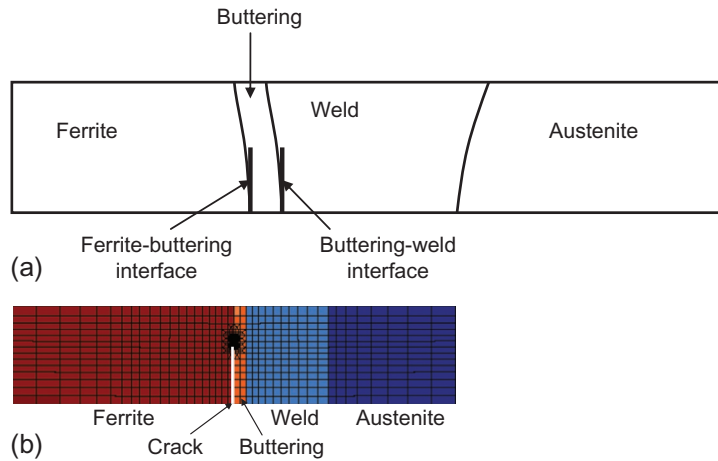
6.9 EFFECT OF SPECIMEN SIZE ON PROBABILITY OF CLEAVAGE FRACTURE IN THE DBTT REGIME

The effect of specimen size on the fracture toughness variation in the DBTT region was also studied. Figure 12.19 presents the predicted results of fracture toughness variation (with temperature and probability of fracture) for the 1T, 2T, and 4T CT specimens, respectively. The 2T and 4T CT specimens are scaled versions of 1T CT specimens with scale factors of 2 and 4, respectively. It can be observed from Figure 12.19 that the scatter in the fracture toughness decreases for the specimens with larger dimensions, signifying the increase in the fracture toughness transition temperature. Hence, specimens with larger dimensions are more susceptible to the DBT for a given value of loading. This is in line with the experimental observations.

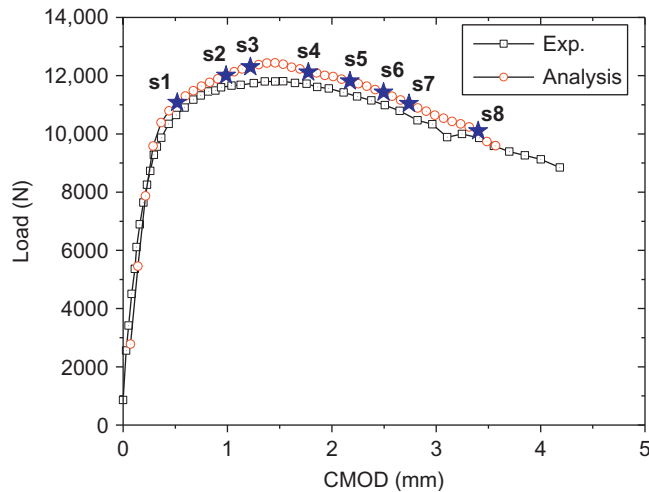
6.10 ANALYSIS OF SEB SPECIMEN WITH CRACK AT THE FERRITE-BUTTERING INTERFACE

The standard dimensions of SEB specimen are same as discussed in Section 5 and these are shown in Figure 12.2a. The schematic representation of crack at the ferrite-buttering interface of the dissimilar metal welded joint is shown in Figure 12.20a. The FE mesh of the SEB specimen with different material properties (Table 12.7) at different regions and initial crack at the ferrite-buttering interface is shown in Figure 12.20b. The full specimen is meshed with FEs as we have different materials at both sides of the crack and hence the symmetric conditions cannot be used.

The load versus CMOD and J -resistance behavior of the SEB specimen is obtained from the analysis and compared with those of experiment in Figures 12.21 and 12.22,

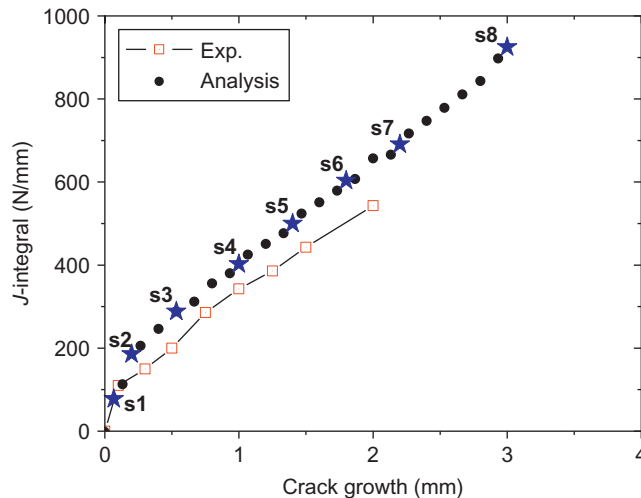
**FIGURE 12.20**

(a) Schematic representation of crack at two different interfaces of the dissimilar metal welded joint, (b) FE mesh of the SEB specimen with different material properties at different regions and initial crack at the ferrite-buttering interface.

**FIGURE 12.21**

Load-CMOD response of the SEB specimen with initial crack at the ferrite-buttering interface (experiment vs. analysis). s1-s8 are the points (star symbols) at which snapshots were taken at different instances of loading.

respectively. As can be seen from these figures, the results of analysis are in excellent agreement with those of experiment and hence, it is demonstrated that the FE formulation with the porous plastic yield function of Rousselier's damage model is an excellent tool for analysis of crack growth and path in dissimilar weld joints. The

**FIGURE 12.22**

J-resistance behavior of the SEB specimen with initial crack at the ferrite-buttering interface (experiment vs. analysis). s1-s8 are the points (star symbols) at which snapshots were taken at different instances of loading.

parameters of individual materials could be successfully used in the weld joint fracture mechanics specimens and hence these are transferable. This method can be used to design a dissimilar weld joint where the parameters like width of buttering and weld layer, material properties of electrodes, etc., can be optimized to obtain a weld with desired mechanical strength and fracture properties (Table 12.7).

7 CONCLUSION

Though local damage models for ductile crack growth simulation have been very successful in predicting the load-deformation and fracture resistance behavior of different types of specimens and components, it is not able to simulate the conditions of large stress gradients at the crack-tip. This is due to the mesh-dependence nature of the solutions. In addition, the size effect cannot be predicted (by scaling the mesh of the reference specimen) unless one changes the mesh design in the analysis. Symmetric boundary conditions are widely used in FE analysis to save computational effort. In certain situations, the symmetric boundary conditions cannot be used due to presence of several types of nonsymmetries in loading, material and boundary conditions. In these situations, the results of the nonlocal model are more realistic when compared to those of local models. Due to the above observation, one must be careful about the use of local damage models for prediction of crack path when the geometry, crack configurations, and loading conditions are complex. The local formulation of the Rousselier model is able to predict the failure behavior of most

specimens and components if the mesh size is coupled to the microstructure and if the above restrictions do not arise in the problem.

The nonlocal model, however, is much more powerful and versatile. The corresponding restrictions with the use of the local formulation are overcome when the nonlocal formulation is used. The results of the nonlocal model are also independent of mesh size. In addition, the nonlocal models are able to accurately predict the scatter in fracture toughness and its variation in the DBTT regime for specimens with various crack-depth, thickness values and sizes. This is due to the ability of the nonlocal models to predict extremely small extents of stable crack growth before the specimen fails by unstable mode of cleavage fracture. In case of the specimen with shallow crack, the fracture toughness transition temperature T_0 as predicted by the nonlocal model is lower compared to that of the deeply cracked specimen and hence, the data from these types of specimens may be more appropriate in structural integrity analysis of components with shallow surface cracks compared to those obtained from standard deeply cracked specimens. Moreover, the nonlocal FE damage model can be reliably used to simulate the fracture behavior of actual components with postulated cracks rather than resorting to carrying out experiments (which may not be feasible in many situations) or using standard deeply cracked specimen data with large values of crack-tip constraints (which may not be representative of the actual crack-tip constraint in the component).

Design and safety analysis of dissimilar weld joints are an important issue for ascertaining structural integrity of these joints which are unavoidable at several locations of thermal and nuclear power plants. In order to optimize the dimensions and properties, one must have a tool to predict the performance of these joints during actual and accidental loading conditions. To prove the safety of these joints, initial cracks of different sizes are usually postulated at different locations, especially the interfaces. Fracture mechanics method cannot be directly used because of difficulties in defining the crack-tip loading parameters. The use of a porous plasticity yield function for initiation and propagation of cracks in ductile materials have several advantages, for example, the material stiffness degrades with progress of damage and simulates a crack (material with negligible stress carrying capability) at the material point with maximum plastic deformation and stress triaxiality. A FE formulation of the Rousselier's damage model has been used in this work to predict the crack path and fracture resistance behavior of a SEB specimen made from the dissimilar weld joint with an initial crack at the ferrite-buttering layer interface. The micro-mechanical parameters have been determined for the individual material and it was demonstrated that they can be safely used to predict the crack growth in the specimen with dissimilar metal welded joint.

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A concise filtergram wear particle atlas and some case studies

13

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1 INTRODUCTION

Wear failure is a common failure in rotary machines and wear particle analysis is an important technique in the wear diagnosis. Wear particle analysis is based on the fact that any wear would generate wear particles (debris). As different wear mechanisms generate different types of wear particles, the wear particles contain critical information regarding the machine health. By analyzing the wear particles, the machinery wear conditions can be assessed. In fact, wear particle analysis is a mature technique with a history of over four decades. Nowadays, wear particle analysis has become a necessary technique in advanced condition monitoring programs.

Wear particle analysis counts on three parameters, i.e., the particle types, sizes, and quantity. Among the three parameters, particle-type identification is crucial. Wear particle identification relies on the particle morphology and surface features. The surface features include colors, texture, and other surface attributes. In terms of wear particle analysis, the particle morphology and the surface features should be taken account.

Broadly speaking, some other methods have also been categorized as wear particle analysis, such as elemental analysis, automatic particle count, and PQ (particle quantify) index. Strictly speaking, as those methods are irrelevant to wear particle morphology, the name of wear analysis, rather than wear particle analysis, is more appropriate to those methods.

To successfully assess the particle type, size, and quantity, ferrography, filtergram (patch test), and computer image analysis constitute the three major techniques of wear particle analysis in modern-day condition monitoring [1].

2 THE WEAR PARTICLE ANALYSIS TECHNIQUES

Ferrography: Invented and developed in 1970s, ferrography ushered in a new era for tribology. Since its invention, wear particles have been extensively studied under the microscope. In ferrography, a magnetic field is applied to separate wear particles from the oil. Since ferrography was successfully applied for first time in wear diagnosis in the US military in 1970s [2], wear particle analysis has been widely applied in various industries from 1980s to 1990s.

There are two main types of ferrography instruments, the analytical ferrography and the DR (direct reading) ferrography. DR ferrography is now obsolete and has almost completely replaced by PQ in industry. In the present day, ferrography generally refers to analytical ferrography.

Analytical ferrography applies a magnetic field to precipitate particles onto a glass slide called a ferrogram. Under the influence of magnetic forces, the magnetized particles are deposited in strings and sorted by their sizes at different positions. The particles on the ferrogram are examined under an optical microscope with bichromatic lights. Green color is applied for the transmission light, while red or white color is for the reflection light. Due to the bichromatic lights simplifying

process of distinguishing particles, less expertise is required in identifying different types of wear particles on the ferrogram. A trained technician can conduct particle identification. Furthermore, ferrogram can be heated or chemically treated for further particle examination.

The main disadvantage of analytical ferrography is the poor capability in detecting the nonferrous particles. Low efficiency in sample preparation is an additional drawback. Due to these disadvantages, ferrogram lacks its popularity in today's condition monitoring. On the other hand, filtergram has some advantages which can compensate for the disadvantages of ferrogram.

Filtergram: Also known as the patch test in the industry, filtergram is actually an old technique. Until the 1990s, filtergram has not been popular in wear particle analysis. Rather than using a magnetic field to deposit particles, filtergram utilizes filter membranes to clog particles, similar to fishing with a net. The filter membrane with the specifications of 1- μm pore size and 47 mm diameter is generally applied in wear particle analysis. Like ferrogram, the particles on filtergram are distinguished under optical microscope by trained technicians. The particles on filtergram can be further studied under SEM/EDS (scanning electronic microscope/energy dispersive spectrometry). In comparison with ferrogram, filtergram has the following advantages and disadvantages [3]:

The advantages in filtergram are:

1. Filtergram clogs all the particles which are larger than the membrane pore size. Hence, either ferrous or nonferrous particles can be all deposited. As a quantitative method, filtergram had been applied in ISO 4407:1991 (the microscopic particle count).
2. In sample preparation, filtergram has much higher efficiency than that of ferrogram.
3. Particle pileup rarely occurs in filtergrams.

The disadvantages in filtergram are:

1. Particle distinguishing requires the expertise. The examination is also time-consuming.
2. For corrosive wear particles, it is difficult to be distinguished. However, such particles are effortlessly spotted via ferrogram, as corrosive particles are piled up at the exit end of the ferrogram.
3. As the filter membrane is opaque, bichromatic microscopes cannot be applied in the filtergram.

Computer Image Analysis: Due to the time-consuming nature of manual microscopic examination, there is a keen desire to handover the task to computers. In last two decades, some efforts have been put in wear particle image analysis [4–12]. The research on particle images aimed to develop an instrument with artificial intelligence. The emergence of the first commercial instrument of computer image analysis in 1998, named LNF (LaserNet Fines), is a milestone in the history of wear particle analysis.

LNF applies laser pulse to count the particles. At the same time, the silhouette images of all particles larger than 20 μm are classified with an artificial neural

network. The artificial neural network was trained by the particle library obtained from ferrographic techniques [12].

As the morphology of the oil degradation particles is diverse, some of these varnish particles share the similar profiles with the metallic wear particles [13], i.e., the pseudo-metal wear particles. Under optical microscope, the pseudo-metal wear particles can mislead even the experts. Hence, computer image analysis would misidentify some varnish particles as fatigue wear particles.

Benefiting from its high efficiency, LNF is used to process large numbers of oil samples for identifying the abnormal wear. However, as recommended by the LNF manufacturers [12], a follow-up technique is needed if LNF detects the abnormal wear. Manual microscopic examination should be carried out to validate the particle classifications. Therefore, the goal of replacing experts with computer image analysis may be realized in the near future.

In fact, filtergram is still a valuable technique in today's wear particle analysis. Different from ferrography, filtergram can display detailed features of individual particles. From this point, a concise wear particle atlas from industrial case studies is presented in this chapter.

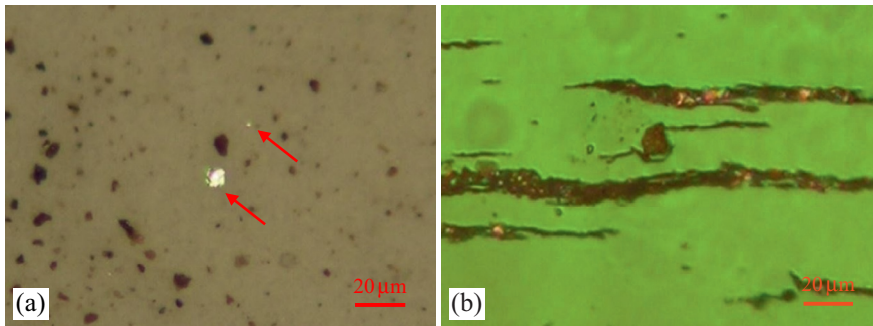
3 RUBBING WEAR PARTICLES

Rubbing wear particles are the most common wear particles, but not important in wear diagnosis. Even in an ideal lubrication condition, such as hydrodynamic lubrication, rubbing wear particles are still generated. It is believed that rubbing wear particles are usually generated from the shear mixed layer (Bielby layer). The shear mixed layer is the metal outer surface on which the atoms are amorphous [14]. Under contact stress, the normal rubbing wear particles detach from the shear mixed layer due to the crystalline structural difference from the bulk metal.

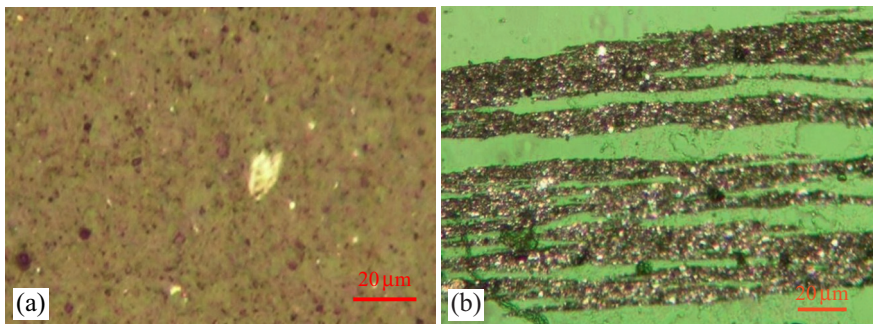
The sizes of the rubbing wear particles range from 0.5 to 15 μm [2]. Figure 13.1a shows individual rubbing wear particles (bright) on filtergram. Some of the small-size particles in the background were the varnish particles (brown). These particles were from an oil sample of a rolling element bearing. The same oil sample was also prepared on a ferrogram. On the ferrogram (Figure 13.1b), the rubbing wear particle formed the strings under magnetic fluxes, whereas most of the varnish particles had been washed away.

In some lubricants with EP (Extreme Pressure) additives, or other additives to form metallic organic compounds, the metallic salt layer is formed on the steel surface under extreme pressure. The tiny wear particles generated from the metallic salt layer resemble rubbing wear particles under optical microscope. Strictly speaking, the particles from metallic salt layer do not belong to the rubbing wear particles.

Figure 13.2a shows the tiny metallic salt layer wear particles on a filtergram. The particles were from a rolling element bearing oil sample with the EP additives. On the filtergram, it is difficult to differentiate the metallic salt layer wear particles from the tiny varnish particles. However, on ferrogram the metallic salt wear particles forms strings (Figure 13.2b) whereas the varnish particles had almost washed away entirely.

**FIGURE 13.1**

The rubbing wear particles from a rolling element bearing oil sample ($200\times$): (a) the rubbing wear particles (bright) and the small-size varnish particles (brown) on the filtergram; (b) the rubbing wear particles forming strings on the ferrogram. Green light was used as the transmission light and bright light as the reflection light in the bichromatic microscope.

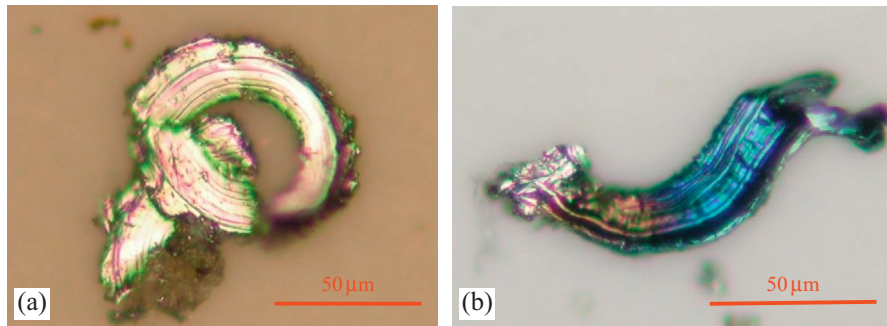
**FIGURE 13.2**

The tiny particles from a bearing oil sample with EP additive ($200\times$): (a) the metallic salt layer wear particles and the varnish particles on the filtergram; (b) the metallic salt layer wear particle on the ferrogram. Green light was used as the transmission light and bright light as the reflection light in the bichromatic microscope.

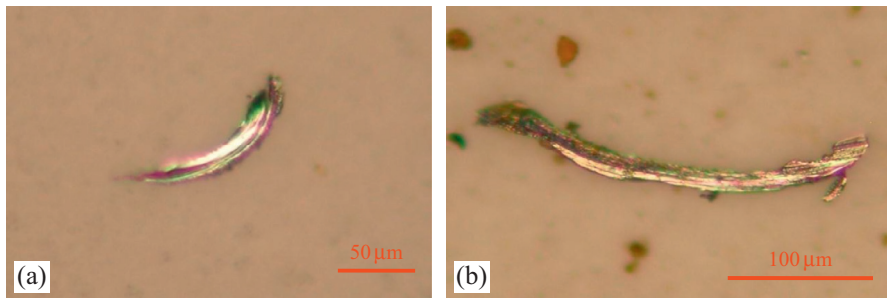
4 CUTTING WEAR PARTICLES

Cutting wear particles are generated from the abrasive wear mode. Resembling the machining swarf under microscope, cutting wear particles are easily differentiated from other particles. There are two types of cutting wear, two-body cutting wear and three-body cutting wear.

The two-body cutting wear is described as one-metal component scraping the opposite component due to misalignment. Morphologically, the two-body cutting wear particles are wider than that of the three-body cutting wear particles.

**FIGURE 13.3**

The two-body cutting wear particles (200 $\times$): (a) a cutting wear particle from a gearbox in cooling tower fans; (b) a cutting wear particle from a gas turbine/generator.

**FIGURE 13.4**

The three-body cutting wear particles (200 $\times$): (a) a cutting wear particle from a hydraulic oil sample; (b) a cutting wear particle from a gearbox.

Figure 13.3 shows the typical two-body cutting wear particles from the oil samples taken from different machines in a 400 Mw thermo-power plant.

The three-body cutting wear is produced by a foreign hard particle cutting on the matching components. The morphology of the three-body cutting wear particles is comparatively slim. Figure 13.4 shows the typical three-body cutting wear particles from oil samples taken from different machines in a paper mill.

5 LAMINAR PARTICLES

Laminar particles are commonly found in the oil samples from rolling element bearings and gearboxes. It is believed that laminar particles are formed by the rolling press over the particles. The major features of laminar particles are a thin, flat surface. Hence, laminar particles are easily distinguished by these features.

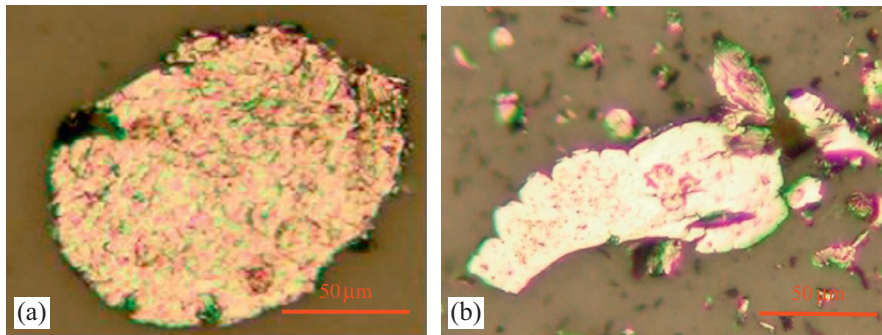


FIGURE 13.5

The laminar particles (200 $\times$) from different machines: (a) a laminar particle from a rolling element bearing in a paper mill; (b) a big laminar particle and the small laminar particles from a gearbox in a pulp plant.

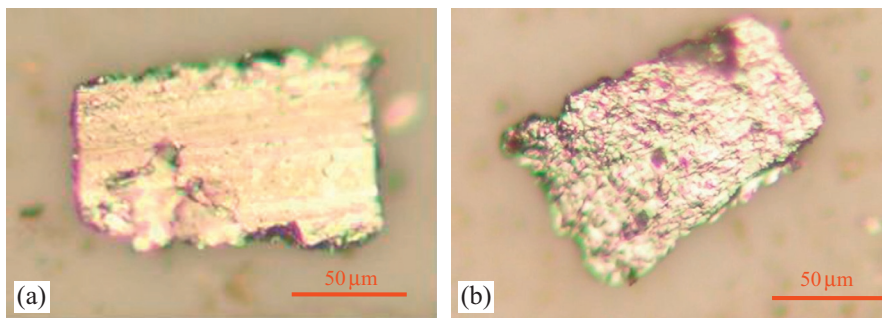
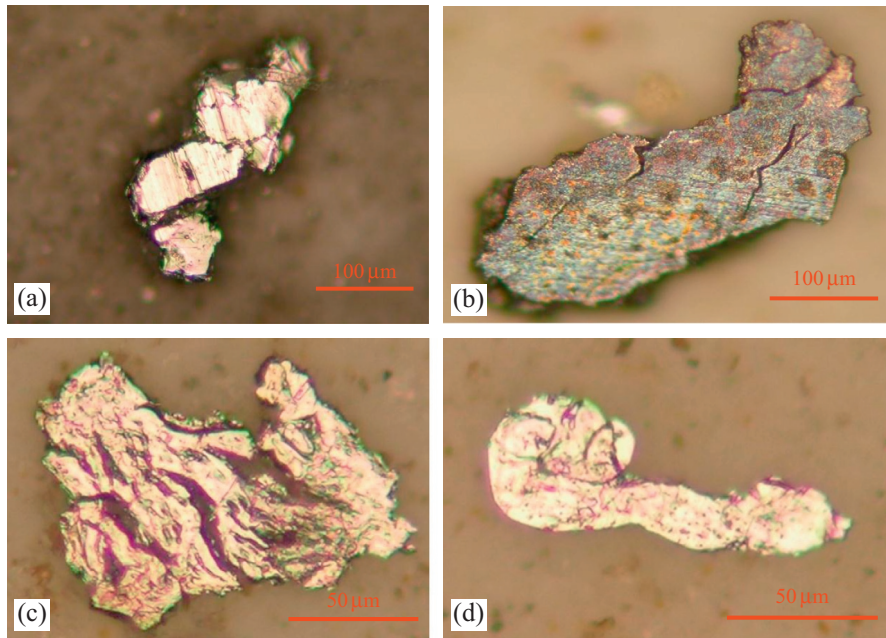


FIGURE 13.6

The front side and the backside of a laminar particles from a rolling element bearing (200 $\times$): (a) the front side with some striations on the smooth surface; (b) the backside with the rough surface.

Figure 13.5a shows a typical laminar particle from an oil sample taken from a rolling element bearing in a paper mill. Figure 13.5 displays the laminar particles from an oil sample taken from a gearbox in a pulp plant.

In filtergram, the front side and the backside of a wear particle can be examined, respectively, with a special technique. Figure 13.6 shows both sides of a laminar particle. This particle was from an oil sample taken from a rolling element bearing. Figure 13.6a shows the front side of the particle. Due to the rolling over, the trace of the striations on the particle surface can still be distinguished. With the magnetic tip, wear particle can be turned over. So the backside of the particle can be revealed. Figure 13.6b is the backside of the particle, showing the rough surface. The rough feature suggests that the backside was from the fatigue crack propagation in the subsurface.

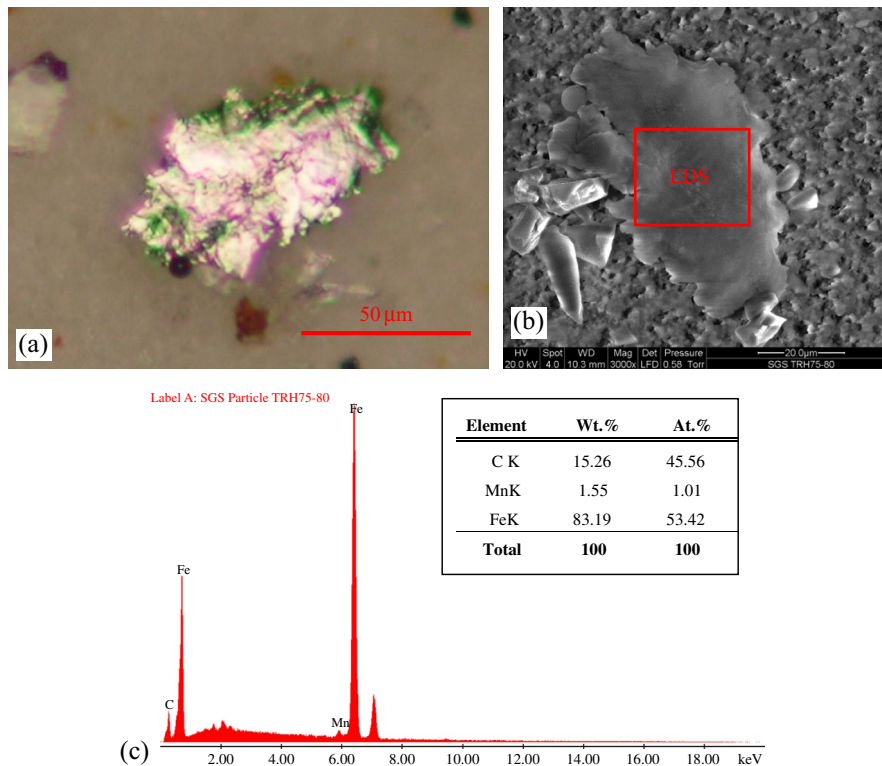
**FIGURE 13.7**

The wear particles had been subjected to different degree of the press (200 $\times$): (a) a fatigue wear particle from a rolling element bearing; (b) a fatigue wear particle from a compressor; (c) a laminar particle subjected severe press; (d) a laminar particle pressed from a cutting wear particle.

During the machine's operation, some wear particles can be further deformed by the machine component's rolling and pressing, forming the laminar particles. Hence, the thickness ratio of laminar particles depends on severity of pressing.

Figure 13.7 displays the wear particles subjected to different degree of press. Figure 13.7a is a fatigue wear particle from an oil sample taken from a rolling element bearing. The particle surface had been flattened. Figure 13.7b is a fatigue wear particle from an oil sample taken from a compressor. The cracks on the flat surface indicate severe press. Both particles remained both features of fatigue wear particle (thicker) and laminar particle (flat surface). Figure 13.7c and d are the wear particles from an oil sample taken from a rolling element bearing in a Yankee machine. In Figure 13.7c, the laminar particle had been pressed so severely, forming the big cracks and the holes. Figure 13.7d is a laminar particle originally from a cutting wear particle.

Some wear particles lose their magnetic properties although the particles are generated from the low-alloy steel components. The mechanisms of magnetic alternation have not been studied. It could be due to the crystal transformation in the particle formation, such as changing from bcc to fcc, or forming amorphous structures.

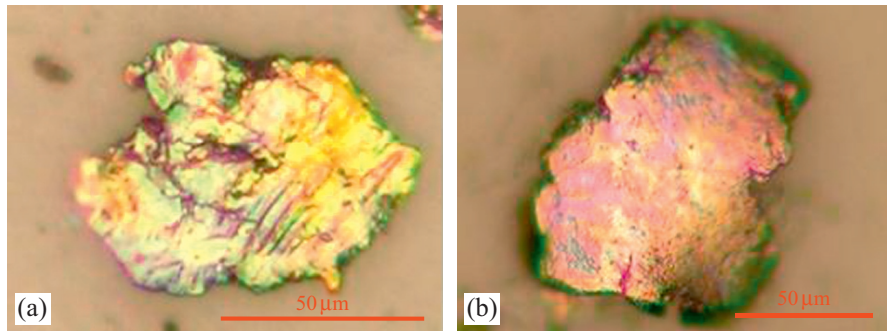
**FIGURE 13.8**

A nonmagnetic laminar wear particle from a wind turbine gearbox: (a) under optical microscope (200 $\times$); (b) under ESEM (3000 $\times$), and (c) EDS analysis results indicating low-alloy steel (1.55% Mn) and the results on non-magnetic laminar wear particle indicating that the particle was from the low alloy steel.

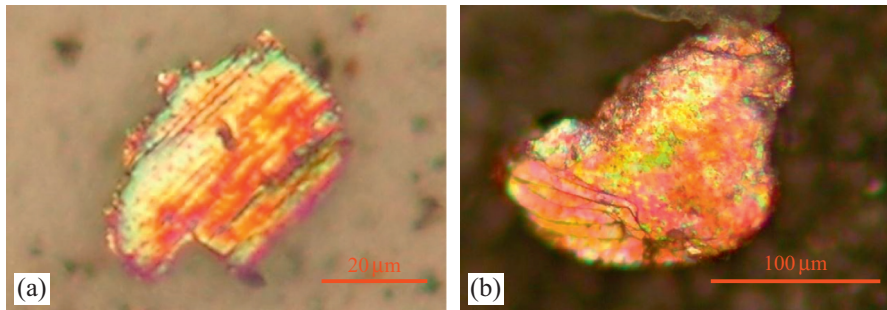
Figure 13.8a is a nonmagnetic laminar particle which cannot be attracted by a magnetic tip. This particle was found in an oil sample from a wind turbine gearbox. *In situ* ESEM/EDS (environmental scanning electronic microscope/energy dispersive spectroscopy) analysis results on the particle (Figure 13.8b) to confirm the low-alloy steel, excluding from austenitic (18-8 type) stainless steels.

During particle formation, oxide film would form on the particle surface due to the heat affect. The oxide films change the particle colors, appearing from straw-yellow to blue. Figure 13.9 shows two laminar particles with the different colors. The particles were from the oil samples taken from different gearboxes in a paper mill.

Generally, red oxide film on the wear particle surface is associated with moisture in the oil [4]. Figure 13.10 displays two laminar particles with red oxide film. Figure 13.10a shows a laminar particle (L50 μm) with the partial red oxide film. Figure 13.10b shows a laminar particle (L180 μm) with partial pink and red oxide film. The particles were from the oil samples taken from different gearboxes in a paper mill.

**FIGURE 13.9**

The laminar particles with different colors (200×): (a) a laminar particle with straw-yellow color; (b) a laminar particle with pink-blue color.

**FIGURE 13.10**

The laminar particles (200×) with partial red oxide surface. (a) a laminar particle with striations and partial red oxide film; (b) a laminar particle with partial red oxide and partial pink oxide film.

CASE 1: THE CATASTROPHIC WEAR IN A RACE CAR GEARBOX

The race car gearbox operates in extremely harsh conditions due to both high speed and the impact loads. The lubricant for the gearbox was a special oil containing the solid EP additive suspension. The solid EP dispersion can cushion gear teeth for preventing tooth breakage and allow the use of the lower viscosities for the high speed.

The special oil had been lubricating the race car gearbox properly. In the racing history, the recorded maximum oil temperature was 94.1 °C. However, another non-solid EP type of gearbox oil was tested in a qualifying racing. The oil temperatures reached to 168 °C during the racing, noticing improper lubrication.

After the qualifying racing, the gearbox was stripped down for an inspection (Figure 13.11a). The gear damages were identified (Figure 13.11b). An oil sample from the failed gearbox had been collected for the analyses. Wear particle analysis identified the catastrophic wear. Figure 13.12 displays some of the catastrophic wear

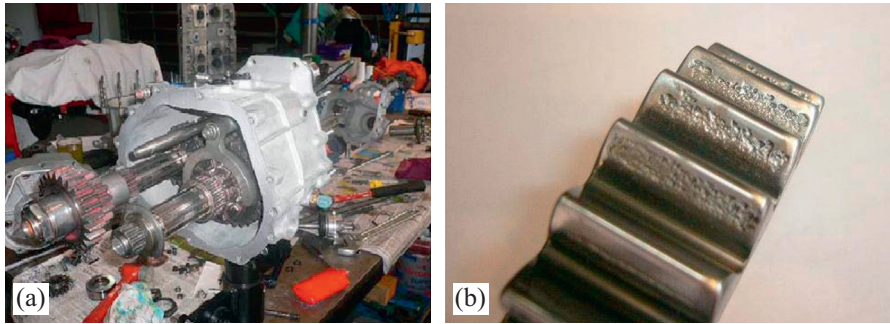


FIGURE 13.11

(a) The stripped down race car gearbox; (b) the damaged gear.

particles in this case study. Due to the inadequate lubrication, the catastrophic wear particles have various oxide films, coloring from straw-yellow to green-blue.

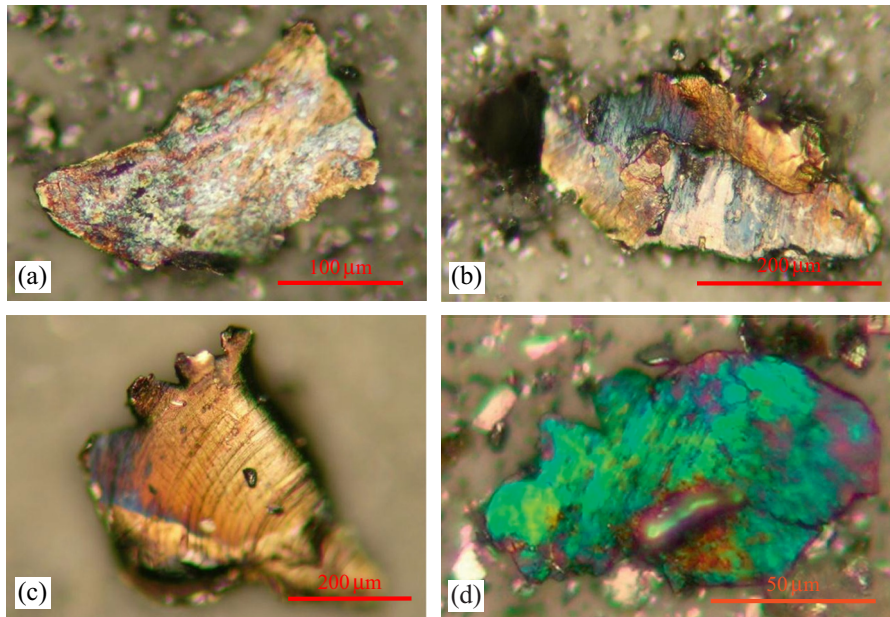
6 FATIGUE WEAR PARTICLES

Fatigue wear particles are generated from the surface contact fatigue. Surface contact includes rolling contact, sliding contact and impact contact. For a rolling element bearing, sometimes, it would subject to the three different contacts which depend on the machine operations. It is believed that microcracks would initiate in the sub-surface after certain cycles of repeat contact stress. When the microcracks propagate to the metal surface, the fatigue wear particles can be generated.

Micropitting and macropitting (spalling) are the forms of surface fatigue failures. Micropitting usually generates the fatigue wear particle up to over a hundred microns. Whereas macropitting produces significant amount of big-size fatigue wear particles, which are up to mm scale. However, such large-size particles are rarely caught in oil samples. But they can be collected by the magnetic plate installed in the oil sump.

Figure 13.13 shows the typical fatigue wear particles. Fatigue wear particles usually feature with smooth surface and random, irregular profiles [2]. In comparison with laminar particles, fatigue wear particles are thick. Figure 13.13a was a fatigue wear particle (L100 μm) from an oil sample taken from a rolling element bearing lube in a Yankee machine. Figure 13.13b was a large fatigue wear particle (L720 μm) from an oil sample taken from a gearbox in a paper mill.

Some fatigue wear particles remain the striations on the particle surface. These striations could associate with the crack propagations during the particle generation. Figure 13.14a was a fatigue wear particle (L160 μm) from the rolling element bearing, on which some striations are distinctive. Figure 13.14b displays another fatigue wear particle from the same bearing. The striations had almost disappeared, but could still be distinguished.

**FIGURE 13.12**

The catastrophic wear particles from a race car gearbox (200 $\times$): (a) a laminar particle with the color oxide film; (b) a laminar particle with the color oxide film; (c) a sliding wear particle with color oxide film; (d) a laminar particle with blue-green color.

Some fatigue wear particles appear as smooth surface on one side and rough surface on another side. [Figure 13.15a](#) was a fatigue wear particle (L140 μm) found in an oil sample taken from a rolling element bearing lube in a Yankee machine. The smooth side with sliding marks ([Figure 13.15a](#)) could be from the rolling-sliding contact surface, whereas the rough side ([Figure 13.15b](#)) could be from the subsurface fatigue.

Oxide film can also form on the fatigue wear particle surface due to the heat affect. [Figure 13.16](#) displays both sides of a fatigue wear particle (L180 μm). This particle was found in an oil sample taken from a rolling element bearing lube in another Yankee machine. As the particle surfaces were rough in both sides, the particle could be generated from the secondary microcrack in the subsurface.

Different color oxide films can form on fatigue wear particles. [Figure 13.17](#) displays the fatigue wear particles with the color variations. In [13.17a](#), the major oxide film was straw-yellow. This particle was from a gearbox oil sample. [Figure 13.17b–d](#) displays the fatigue wear particles from the roller element bearings in the different Yankee machines. [Figure 13.17b](#) was a fatigue wear particle with partial dark blue and partial yellow; [Figure 13.17c](#) was a fatigue wear particle (L650 μm) with partial red oxide and partial dark oxide; [Figure 13.17d](#) was another fatigue wear particle (L320 μm) with dark oxide.

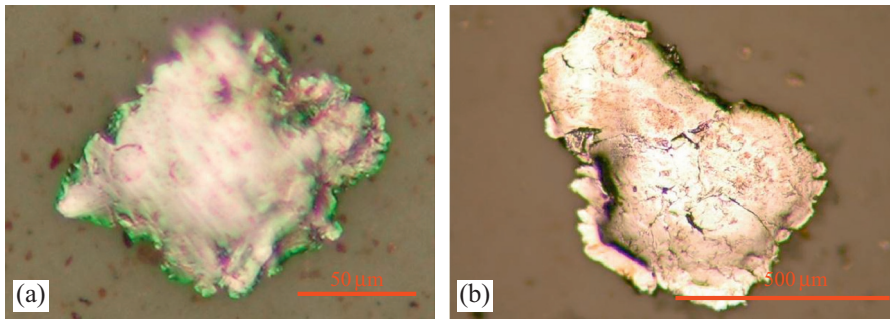


FIGURE 13.13

The typical fatigue wear particles (200 $\times$): (a) a fatigue wear particle from a rolling element; (b) a fatigue wear particle from a gearbox.

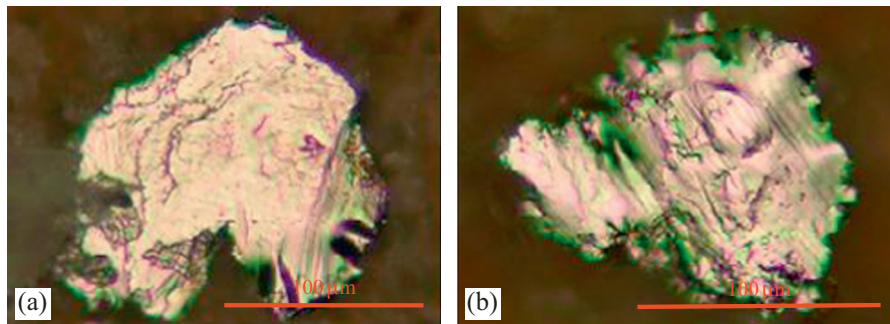


FIGURE 13.14

The fatigue wear particles remaining the striations (200 $\times$): (a) a fatigue particle from rolling element bearing; (b) a fatigue wear particle from the same bearing.

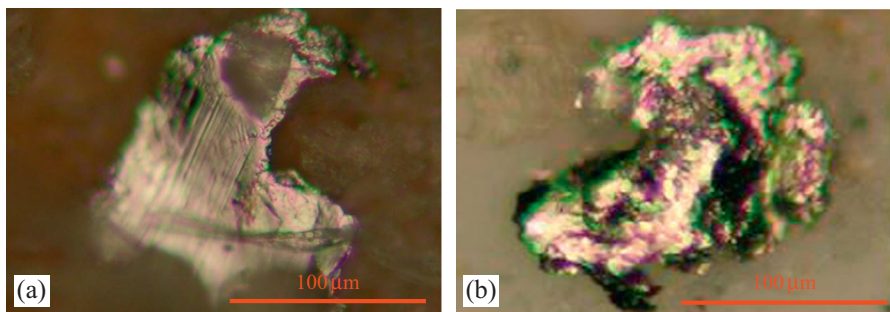
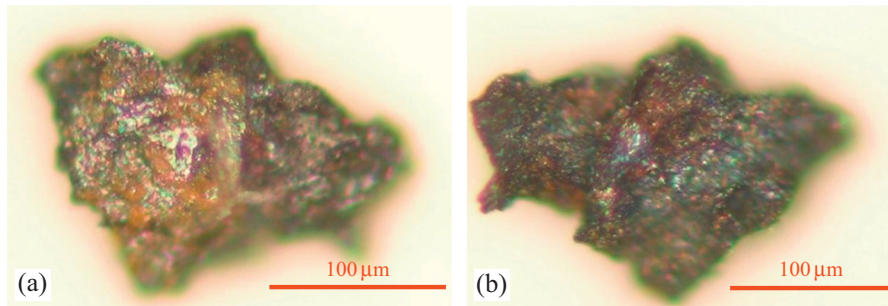


FIGURE 13.15

The front side and the backside sides of a fatigue wear particle from a rolling element bearing (200 $\times$): (a) the front side with the striations on the smooth surface; (b) the back side with comparative rough surface.

**FIGURE 13.16**

The front side and the back side of a fatigue wear particle (200 $\times$): (a) the front side; (b) the back.

CASE 2: THE WIND TURBINE GEARBOX PREMATURE FAILURES

Wind turbine gearbox premature failures are a tough technical problem worldwide [15,16]. The gearboxes designed for 20 years of service life, often suffers failures in between 2 and 11 years [17], mainly occurring in first stage of the planet gear bearings. The failure mode was spalling. The culprit of the spalling was the white etching cracks (WEC), a kind of microstructural alternation.

Figure 13.18 displays the micropitting and the macropitting on an overhauled bearing from a wind turbine gearbox. This bearing failed in only 3 years of operations. Figure 13.19a shows the cross section of the inner ring's microstructures under optical microscope, in which the WEC were apparent. Under ESEM, the spalling by the microcracks is displayed (Figure 13.19b).

In the process of spalling, significant amount of large fatigue wear particles were generated. Figure 13.20 shows the typical wear particles from a wind turbine gearbox.

In the fatigue crack propagation, some striations can be left on the fatigue wear particle surface. Figure 13.21a shows a large fatigue wear particle (L870 μm) under stereomicroscope (45 $\times$). The striations can be distinguished. Under the high resolution and deep focus of ESEM, the striations can be clearly displayed (Figure 13.21b).

7 SEVERE SLIDING WEAR PARTICLES

Severe sliding wear particles are generated from adhesive wear mode. In the excessive stress or in the improper lubrication conditions, the lubrication region would breakdown. The metal to metal contact would generate significant amount of severe sliding wear particles. Generally speaking, adhesive wear is a catastrophic wear mode. In extreme cases, the components would seize.

Sliding wear particles feature with the surface striations formed by the sliding metal contact. As the adhesive wear deteriorates, the striations enlarge and deepen. If the cold weld occurs, the particle surface features would appear to be torn from the adhesive.

Figure 13.22 displays the severe sliding wear particles from different machines. Figure 13.22a and b were the severe sliding wear particles from Case Study 2.

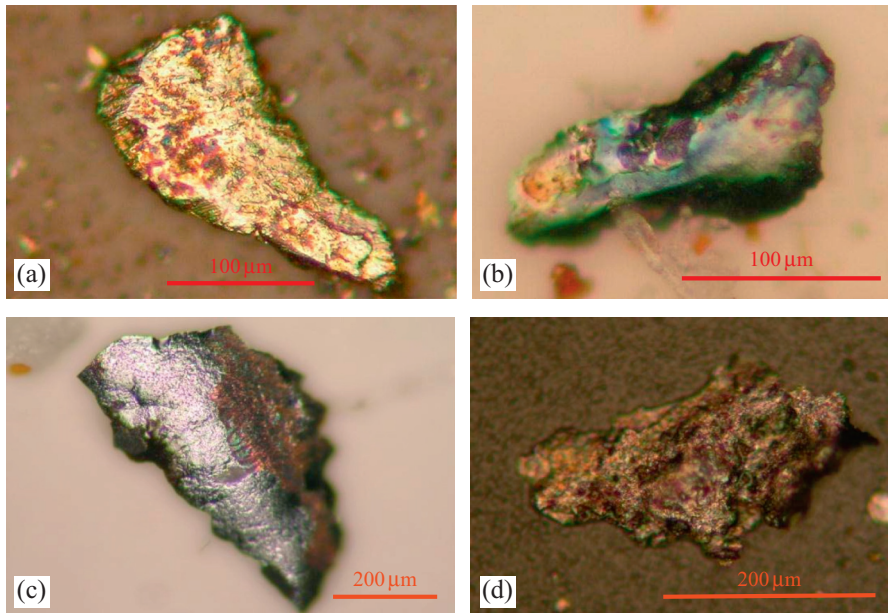


FIGURE 13.17

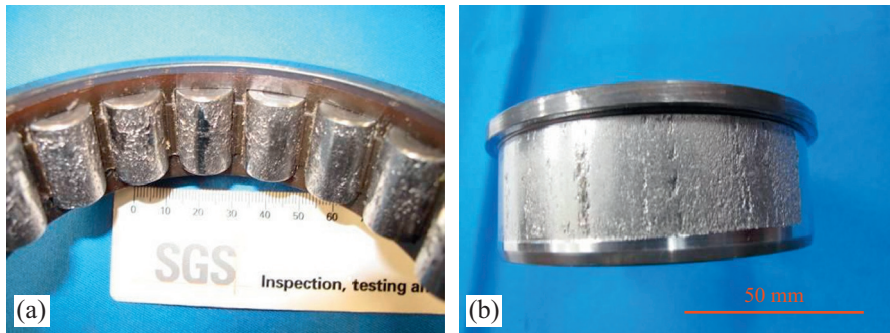
The fatigue wear particles with different color oxide films (200 $\times$): (a) a fatigue wear particle with yellow oxide film; (b) a fatigue wear particle with partial dark blue oxide film; (c) a fatigue wear particle with partial dark oxide film and partial red oxide film; (d) a fatigue wear particle with dark oxide film.

Figure 13.22c was a severe sliding wear particle from the Case Study 1. The big and deep striations with the feature of torn from adhesive indicate the severe adhesive wear. Figure 13.22d was another severe sliding wear particle from an oil sample taken from a rolling element bearing lube in a Yankee machine. The deep striations and the blue color suggest high temperatures were *in situ*.

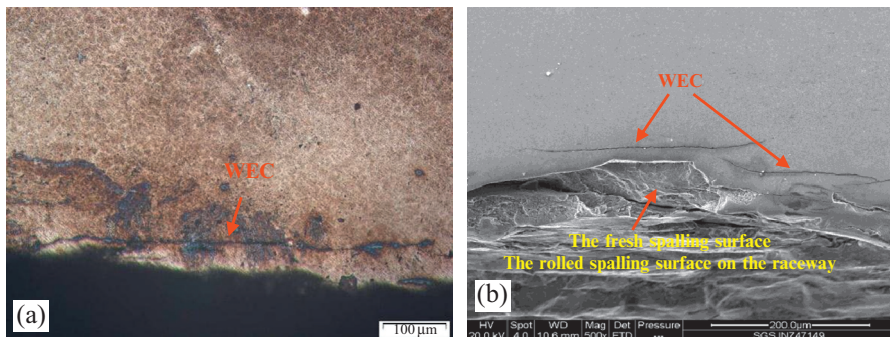
CASE 3: THE SEIZURE FAILURES OF A BULLDOZER DIESEL ENGINE

A bulldozer diesel engine suffered a seizure failure. The bulldozer owner suspected that the engine was deliberately sabotaged by a third party, so an oil sample was taken from the engine oil sump for an investigation. Visual inspection found 70% of free water in the oil sample. However, oil analyses excluded that the water was from the coolant leaking. Wear particle analysis uncovered the catastrophic wear.

Figure 13.23 displays the catastrophic wear particles from the oil sample. As the lubrication region was destroyed by the water, adhesive wear occurred. Figure 13.23a and b were the severe sliding wear particles, featuring the deep striations on the particle surfaces. Figure 13.23c was a fatigue wear particle (L180 μm) with partial straw-yellow oxide film. Figure 13.23d was a copper alloy fatigue wear particles with the colorful oxides. The oxides on the particle surfaces indicate the high *in situ* temperatures during particle generation.

**FIGURE 13.18**

The micropitting and macropitting on a first-stage planet bearing overhauled from a wind turbine gearbox: (a) the rollers; (b) the inner ring.

**FIGURE 13.19**

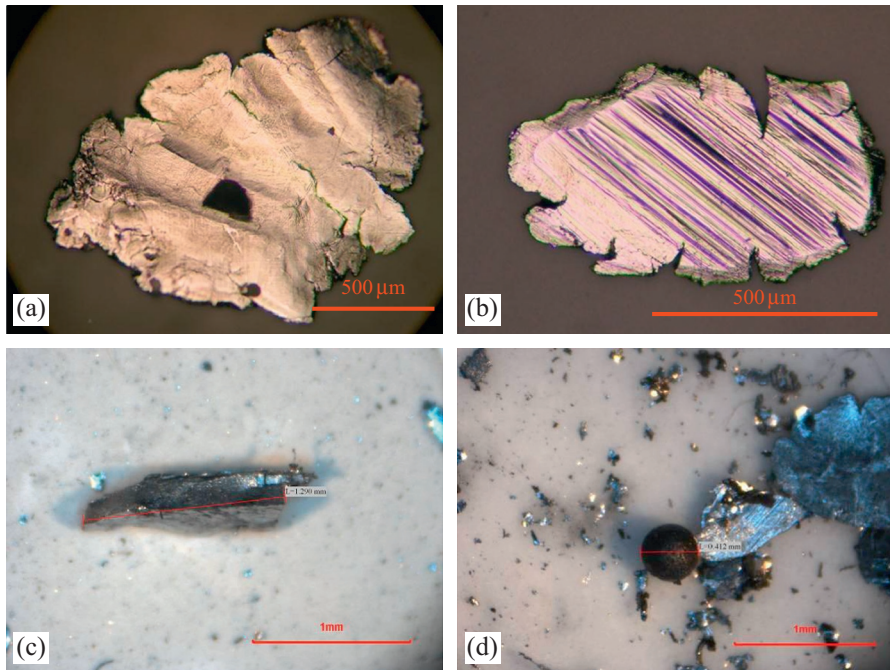
The microstructures of the above bearing inner ring: (a) showing the white etching cracks (WEC) in the subsurface under optical microscope (200 $\times$); (b) the formation of spalling as a result of the WEC propagation (ESEM 500 $\times$).

8 THE OXIDE WEAR PARTICLES

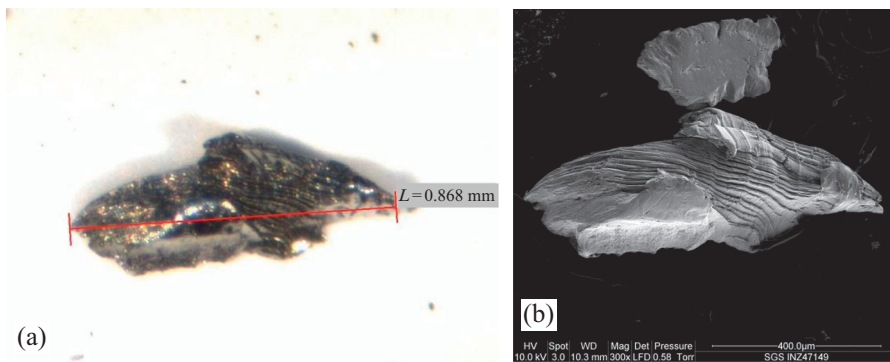
Oxide wear particles are referred to the wear particles with oxide scale. There are two types of oxide scales, red oxide scales and black oxide scales. Red oxide scale is associated with the moisture in the oil, whereas black oxide scale is believed to be associated with high temperatures due to the improper lubrication.

Iron oxides have different crystals and different colors. Hematite ($\alpha\text{-Fe}_2\text{O}_3$) and magnetite (Fe_3O_4) are the most common in nature. Red oxide is believed as hematite ($\alpha\text{-Fe}_2\text{O}_3$) [2], which is similar to “rust.” For black oxide wear particles, their oxide scale has not been studied.

Figure 13.24 shows the typical red oxide wear particles. Both the particles were found in an oil sample with the extremely high moisture content (24,660 ppm). The oil sample was taken from a forward drive gearbox in a paper mill.

**FIGURE 13.20**

The typical large wear particles from a failed wind turbine gearbox: (a) a laminar particle (200 $\times$); (b) a sliding wear particle (200 $\times$); (c) a fatigue wear particle under stereomicroscope (45 $\times$); (d) a spherical particle under stereomicroscope (45 $\times$).

**FIGURE 13.21**

A large fatigue wear particle (L870 μm) from a failed wind turbine gearbox: (a) under stereomicroscope (45 $\times$); (b) under ESEM (300 $\times$).

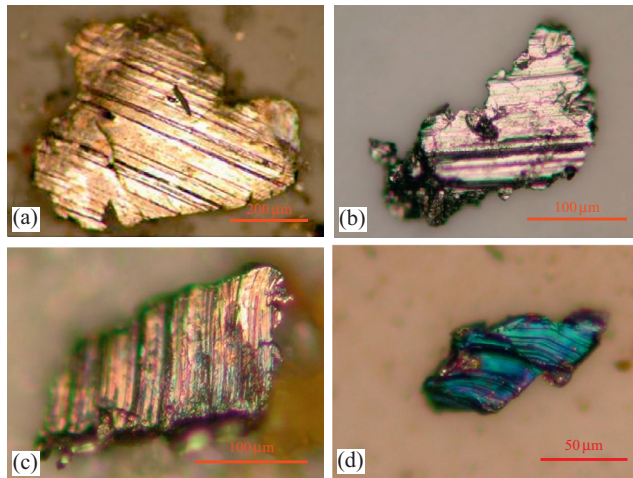


FIGURE 13.22

The severe sliding wear particles with different color oxide film (200 ×): (a) a severe sliding wear particle with dark oxide; (b) a severe sliding wear particle with black oxide; (c) a severe sliding wear particle with the straw and dark oxide; (d) a severe sliding wear particle with the blue oxide.

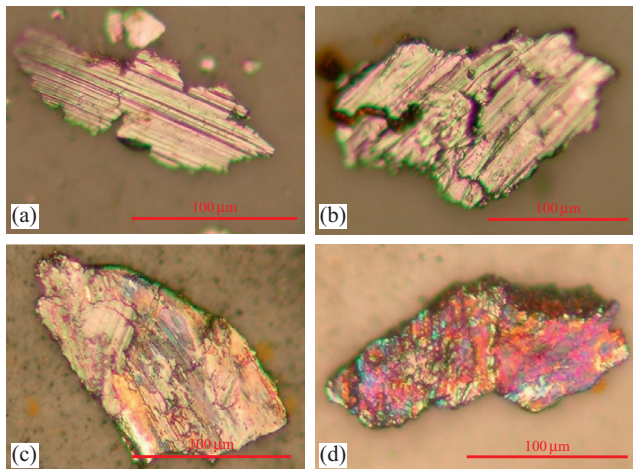


FIGURE 13.23

The catastrophic wear particles from an oil sample taken from a seized diesel engine (200 ×): (a) a severe sliding wear particle; (b) a severe sliding wear particle; (c) a fatigue wear particle with partial straw-yellow oxide film; (d) a copper alloy fatigue wear particle with colorful oxide film.

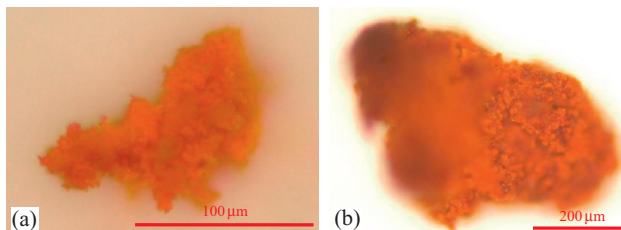


FIGURE 13.24

The red oxide wear particles from a gearbox with the extremely high moisture content (200 ×): (a) a red oxide particle; (b) a red oxide particle.

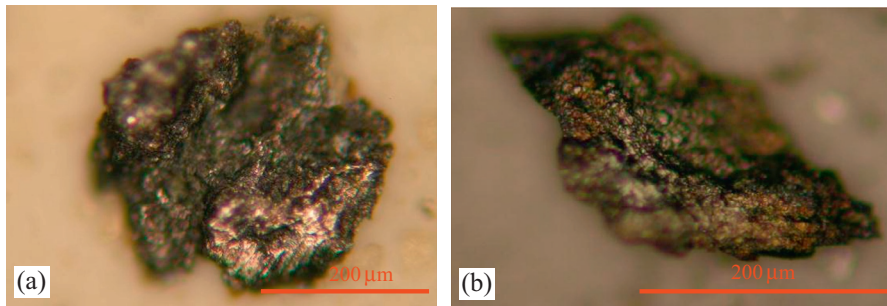


FIGURE 13.25

The black oxide particles from different machines (200 $\times$): (a) a black oxide particle from an air compressor; (b) a black oxide particle from a roller bearing.

Figure 13.25 shows the typical black oxide wear particles from the oil samples with normal moisture contents. The particle in Figure 13.25a was from an oil sample taken from an air compressor. Figure 13.25b was a particle from an oil sample taken from the return line of a roller bearing in a Yankee machine.

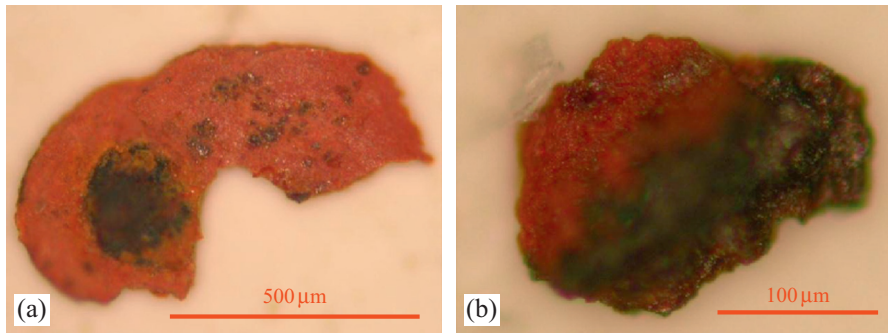
The influence of the moisture on the particle oxidation is complex. Generally, the moisture forms red oxide scale on the wear particle surface, changing the particle's colors from orange to dark brown. On the other hand, the improper lubrication would form black oxide scale. Hence, the multi-oxide scales can form on the single-particle surface. Figure 13.26 shows the wear particles with mixture of red oxide scale and black oxide scale. These two particles were from the same oil sample with extremely high moisture content (Figure 13.24).

In high moisture condition, the red oxide crystals (like the rust) can further grow on wear particle surfaces. Figure 13.27 shows the red oxide crystals on the wear particle surface. Under polarized reflected light, the red oxide crystals remain with their colors and brightness (Figure 13.27b).

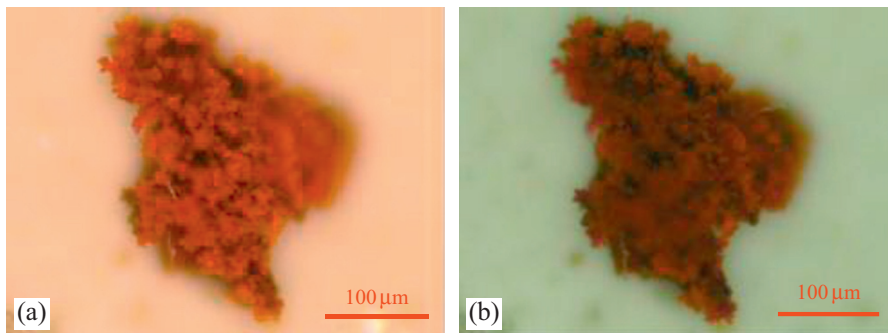
In lubricating oil with extremely high moisture contents, a milky oxide layer can be formed on the wear particle surfaces. Figure 13.28 displays wear particles covering with milky layers. These particles were from two different gearboxes in a paper mill. Both oil samples contained extremely high moisture contents due to the seal failures. In oil sample of Figure 13.28a, 25% free water was found in the oil sample bottle. In oil sample of Figure 13.28b, the moisture content was over 10,000 ppm. The nature of the milky layer has not been studied. It could be iron hydroxide or the oil degradation products generated by the extremely high moisture.

CASE 4: THE WEAR PARTICLE WITH MULTI-OXIDE SCALES FROM AN OIL SAMPLE WITH HIGH MOISTURE CONTENT

In a paper mill, a main lubricant tank (3000 l capacity) supplied the oil to lubricate the two major roller bearings in a Yankee machine, and also other rolling element bearings in the production line. In any event of seal failure, water would infiltrate

**FIGURE 13.26**

The particles mixture oxide scales (200 $\times$): (a) a particle with partial red oxide scale and partial black oxide scale; (b) a particle with partial black oxide scale and partial red oxide scale.

**FIGURE 13.27**

The red oxide crystals (like the rust) growth on a wear particle surface (100 $\times$): (a) under white reflected light; (b) under polarized reflected light.

the lubrication system. Hence, routine oil analysis was a part of the condition monitoring program.

When the Karl Fisher method detected the high level of moisture content (3970 ppm) in a sample, various color oxide wear particles had also been found in the oil sample. Figure 13.29 shows the fatigue wear particles with multi-oxide scales, coloring from yellow to dark brown.

Some tiny red oxide particles ($L < 10\ \mu\text{m}$) had also been found in filtergram (Figure 13.29). Under the polarized reflected light, the red oxide particles remain their colors and brightness (Figure 13.30b). These tiny red oxide particles could be broken down from the red oxide scale which had been fully oxidized.

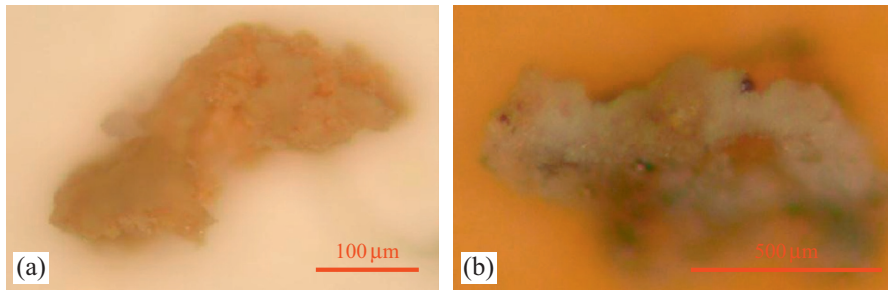


FIGURE 13.28

The wear particles covering with milky layers (200 $\times$): (a) a fatigue wear particle from a gearbox lube with 25% free water; (b) a fatigue wear particle from another gearbox lube with extremely high moisture content (> 10,000 ppm).

9 THE SPHERICAL PARTICLES

Spherical particles are often found in fatigue failures of rolling element bearings and gears. It is generally believed that the spherical particles are formed from molten metals. Like the water bead in nature, the formation of the spherical particles is to minimize surface energy. [Figure 13.31](#) shows a partial molten wear particle found in Case Study 2. [Figure 13.31a](#) is the particle under stereomicroscope. [Figure 13.31b](#) reveals the particle details under ESEM. As the wear particle was partially molten, the top portion appears the water bead feature whereas the bottom portion retains the unmolten features. This evidence confirms that the *in situ* temperature can be so high in some particles.

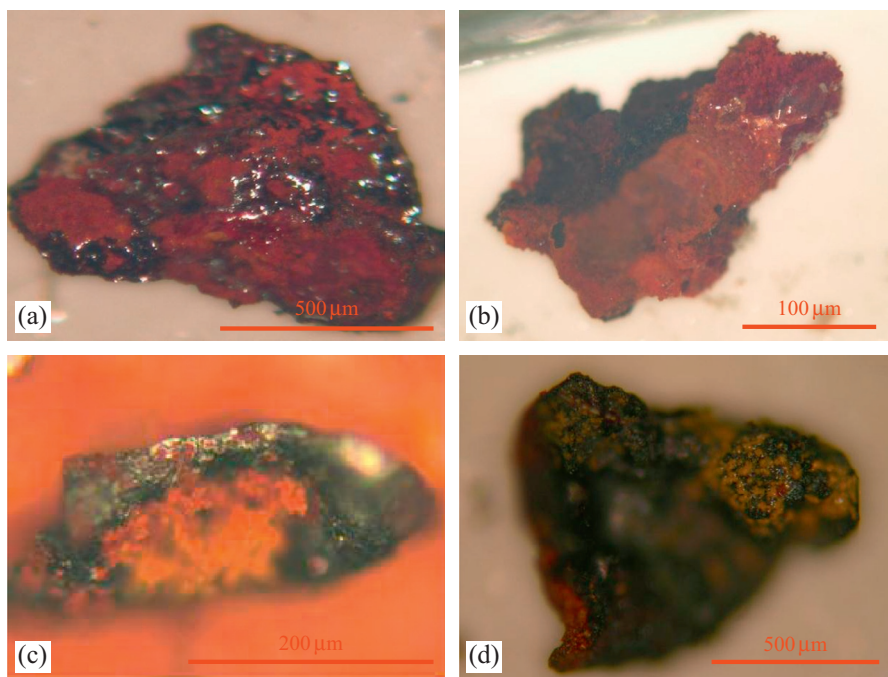
The diameters of spherical particles can range from 5 μm to several hundred microns. In oil samples, the spherical particles from 10 to 50 μm are commonly found. [Figure 13.32a](#) is a spherical particle (D100 μm) under stereomicroscope (45 $\times$). The particle was from the Case Study 2. [Figure 13.32b](#) shows another spherical particle (D50 μm) under metallographic microscope. The particle was from an oil sample taken from a rolling element bearing lube.

The formation of spherical particles is not limited to steels. Brass and bronze components can also generate spherical particles. [Figure 13.33](#) is a bronze spherical particle found on an oil filter of the journal bearings of a geothermal steam turbine.

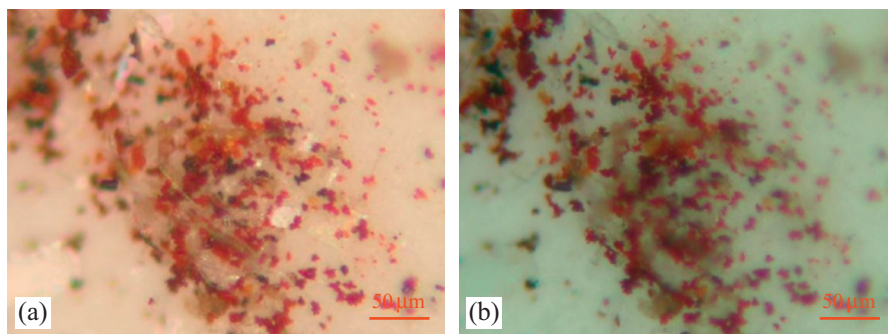
In addition to metals, oil degradation products can form spherical particles. [Figure 13.34](#) shows the two varnish spherical particles with different colors. The particles were from two different hydroturbine oil samples.

10 COPPER ALLOY WEAR PARTICLES

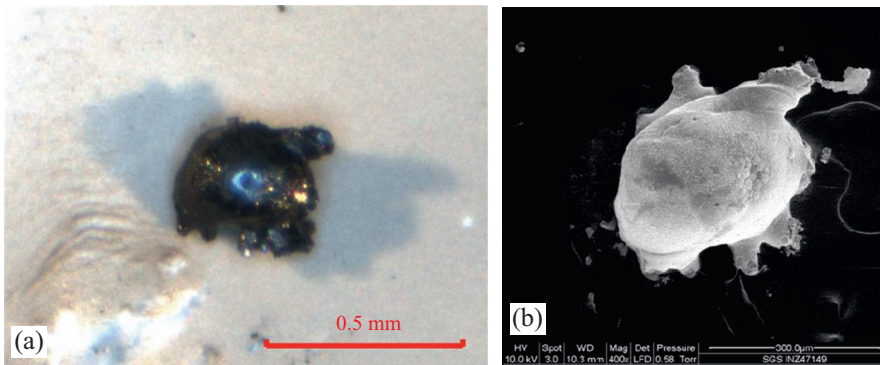
Copper alloy, particularly bronze, is widely applied in the machinery components, such as bearing cage, worm gear, and some slide bearing pads. Hence, copper alloy wear particles are not uncommon on filtergrams. Generally, copper alloy wear particles are easily recognized by their reddish-yellow color.

**FIGURE 13.29**

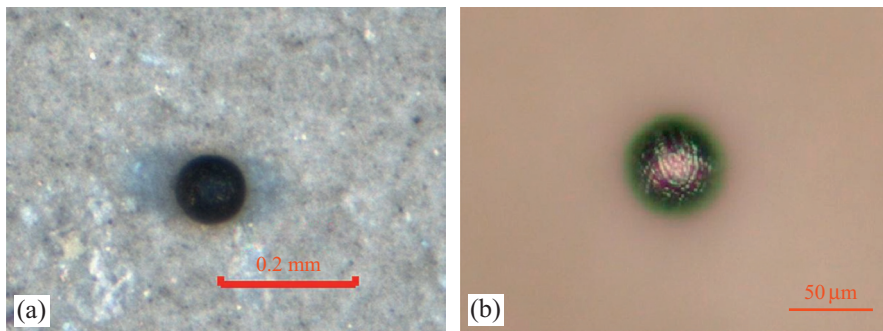
The fatigue wear particles with multi-oxide scales from an oil sample with the high moisture content: (a) a fatigue wear particle (50 $\times$); (b) a fatigue wear particle (200 $\times$); (c) a fatigue wear particle (200 $\times$); (d) a fatigue wear particle (50 $\times$).

**FIGURE 13.30**

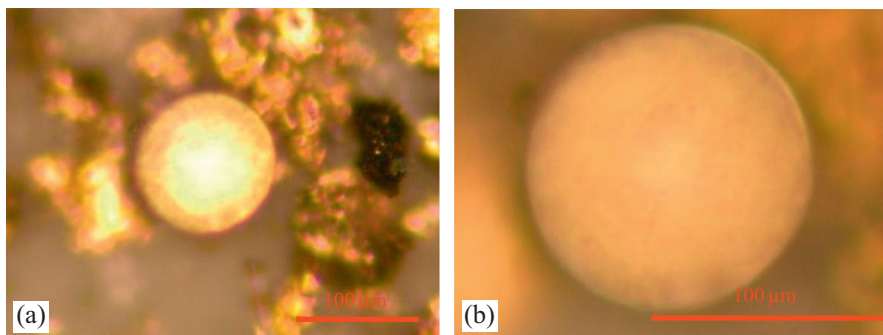
The minute red oxide particles (200 $\times$): (a) under white reflected light; (b) under polarized reflected light.

**FIGURE 13.31**

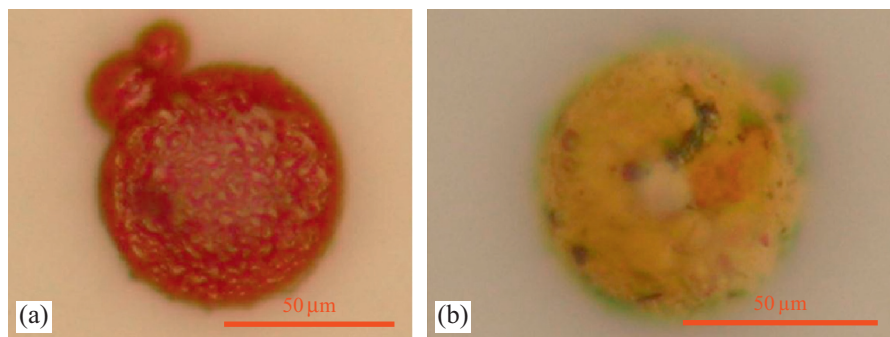
A partial molten wear particles from the Case Study 2: (a) under stereomicroscope (45 $\times$); (b) under ESEM (400 $\times$).

**FIGURE 13.32**

The spherical particles: (a) a spherical particles from the Case Study 2 (stereomicroscope 45 $\times$); (b) the spherical particles under metallographic microscope (100 $\times$).

**FIGURE 13.33**

A bronze spherical particle found in an oil filter of the journal bearings of a geothermal steam turbine: (a) (100 $\times$); (b) (200 $\times$).

**FIGURE 13.34**

The varnish spherical particles from a hydro turbine oil samples (200 $\times$): (a) a brown varnish spherical particle; (b) a yellow varnish spherical particle.

Figure 13.35 shows the typical copper alloy wear particles. Figure 13.35a is a copper alloy fatigue wear particle (L240 μm). The particle could be from the bearing cage. Figure 13.35b is a copper alloy sliding wear particle (L100 μm) from an oil filter in a geothermal steam turbine.

The oxide film can form on copper alloy wear particles by the heat affect. After the oxidation, copper alloy wear particles appear different colors. Figure 13.36a–d displays the copper alloy wear particles with different oxide films, coloring from dark-brown to green.

The light-green copper alloy wear particle (Figure 13.36d) is a particle was from an oil sample taken from the oil return line of the generator babbitt bearing. So it was misconceived as babbitt alloy wear particle with oxide film. However, ESEM/EDS analysis results (Figure 13.37) turned out to be a brass wear particle.

CASE 5: THE IMPENDING CATASTROPHIC WEAR IN A GEARBOX (250 Kw) DUE TO THE OIL STARVATION

A three-stage reducer gearbox (250 Kw) in a paper mill (Figure 13.38a) suffered an impending catastrophic wear due to the oil starvation. In the gearbox, lubrication oil was supplied by the pipelines to each gear and bearing. When the routine wear particle analysis detected the wear issue, an emergency shutdown was implemented for the inspection. Upon the opening the inspection window in the gearbox, the problem was revealed; the pipeline supplying the oil to the side bearing (Figure 13.38b) had been blocked by pulp fibers. Hence, severe wear occurred in the bearing. Owing to the wear particle analysis, the impending catastrophic wear was remedied in a timely manner.

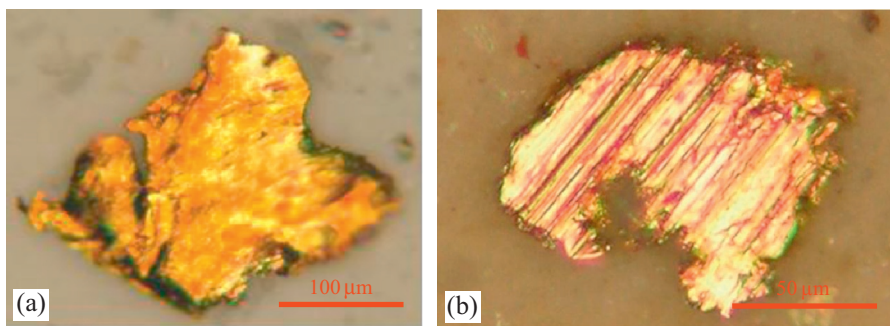


FIGURE 13.35

The copper alloy wear particles (200 $\times$): (a) a copper alloy fatigue wear particle from a gearbox; (b) a copper alloy sliding wear particle from a geothermal steam turbine.

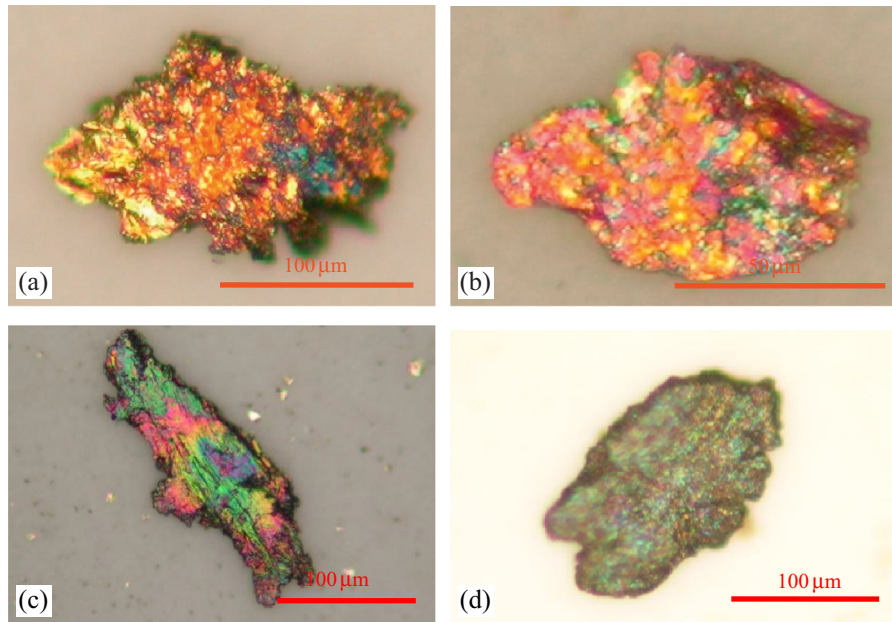
Figure 13.39 shows the typical severe wear particles in this scenario. The particles were from an oil sample taken from the gearbox drain. The copper alloy sliding wear particle (Figure 13.39a) could be from the bearing cage. Other large-size wear particles (Figure 13.39b–d) signify the onset of severe wear.

11 BABBITT ALLOY WEAR PARTICLES

Babbitt alloys (commonly known as “white metal” in the industry) are widely applied in the slide bearing pads due to its low coefficient of friction, high fatigue strength, and excellent seizure resistance. Babbitt alloys are tin-base alloys with Sn-Sb-Cu and Sn-Sb-Pb systems commonly applied to bearing pads.

Unlike steel wear particles, large-size babbitt alloy wear particles are rarely caught in the oil samples. But they can be caught from the oil drain or found on the oil filters. Figure 13.40 shows the large-size babbitt alloy wear particles from the oil drain of a hydro-turbine. The particle in Figure 13.40a retains the “white metal” appearance. However, the tin-base alloys are sensitive to oxidation. Babbitt alloy wear particles usually appear with colorful oxide film on the particle surface (Figure 13.41b). Figure 13.41 shows the babbitt alloy wear particles with different color films.

After oxidation, some color babbitt alloy wear particles would be misinterpreted as other types of particles. For example, Figure 13.42a looks like a copper alloy wear particle, whereas Figure 13.42b resembles a steel wear particle. However, both particles were from the hydro-turbine. ESEM/EDS analyses results (Figures 13.43 and 13.44) of the particles further confirm that both the two particles were from the Sn-Sb-Cu babbitt alloys.

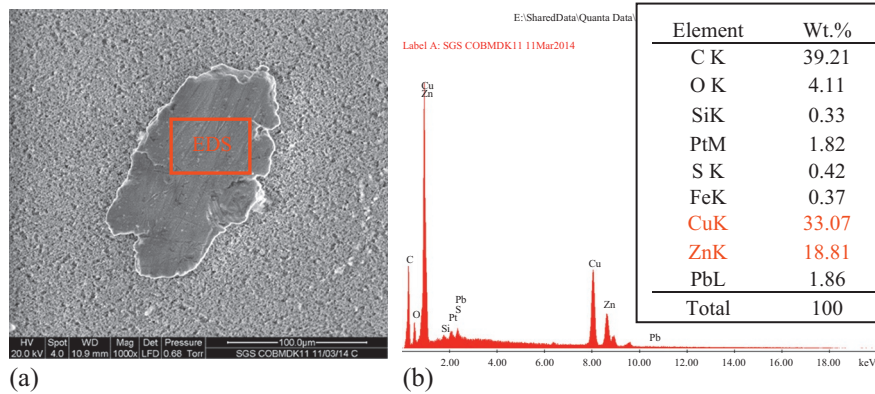
**FIGURE 13.36**

The copper alloy wear particles with different colors (200 $\times$): (a) a copper alloy wear particle from geothermal steam turbine; (b) a copper alloy wear particle from Case Study 5; (c) a copper alloy wear particle from a compressor journal bearing; (d) a light-green copper alloy wear particle from the oil return line of a generator babbitt bearing.

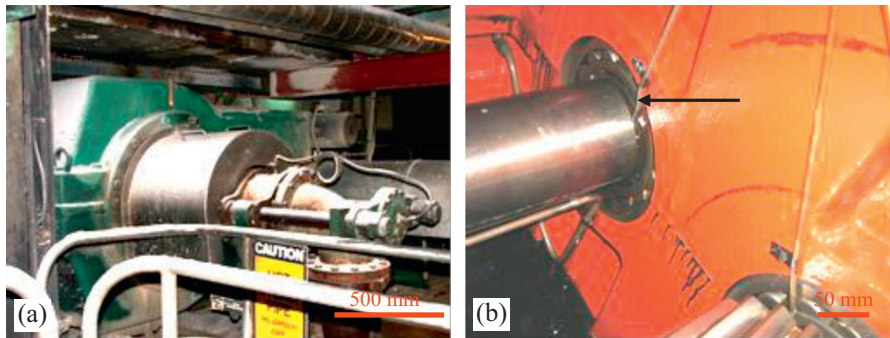
Figure 13.45 shows the typical babbitt alloy wear particles without the oxide film. These particles were from an oil filter of a geothermal steam turbine-generator. The background particles were the bronze wear particles. As the bearing was a multilayer bearing, significant amount of bronze particles had also been generated from the bronze layer.

12 OIL DEGRADATION PARTICLES

Oil degradation occurs by oxidation, nitration, and thermal cracking. Oil oxidation forms carboxylic acids, free radicals, aldehydes and ketones, esters, etc. [26,27]. These oil oxidation products can further react to form the oil degradation particles. It is believed that oil nitration is the major cause of the varnish (lacquer) buildup [17]. The compositions of the oil degradation particles are very complicated. In term of particle appearance, oil degradation particles can be roughly divided into three major categories: friction polymer, varnish, and additive degradation.

**FIGURE 13.37**

The ESEM/EDS analysis results on a brass wear particle: (a) the particle ESEM image (1000 $\times$); (b) the EDS analysis results.

**FIGURE 13.38**

(a) The three-stage reducer gearbox (250 KW) in a paper mill; (b) inside the gearbox—the blockage of the oil supply pipeline being indentified.

12.1 FRICTION POLYMER

Friction polymer particles normally appear semi-translucent under optical microscope. Figure 13.46a is a friction polymer particle from a new oil sample. Some of the friction polymer particles appear different colors. Figure 13.46b is colorful friction polymer particle from a gas turbine oil sample. This particle partially remains the semi-translucent features.

Friction polymer fiber particles are not uncommon on filtergram. Figure 13.47a is a red color friction polymer fiber particle from gearbox oil. In the background, there are some glass fiber particles. Friction polymer fiber particles can be differentiated from cloth fiber particles and glass fiber particles. Glass fiber particles are

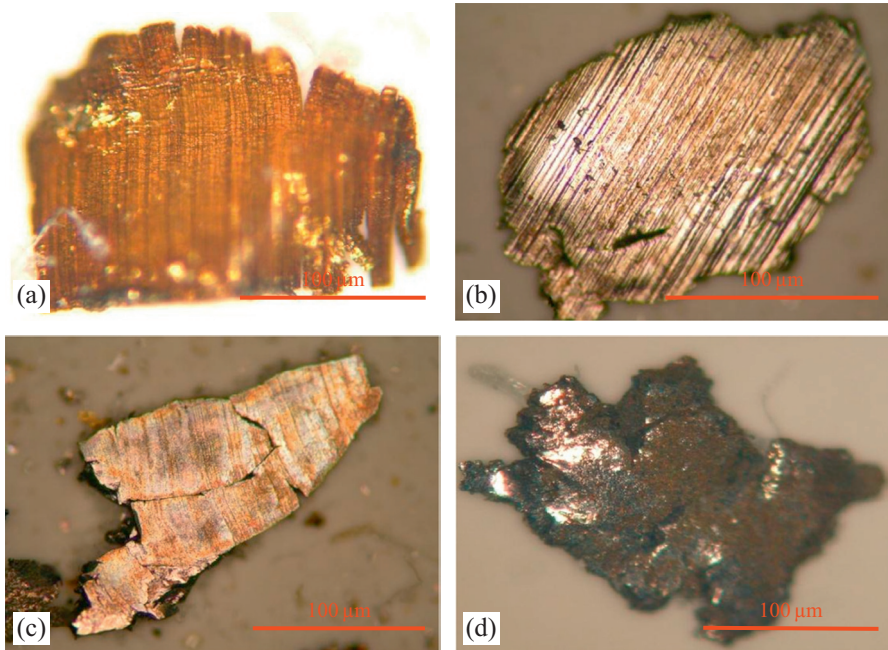


FIGURE 13.39

The severe wear particles from an oil sample taken from the gearbox drain (200×): (a) a copper alloy sliding wear particle; (b) a sliding wear particle; (c) a fatigue/laminar wear particle; (d) a black oxide fatigue wear particle.

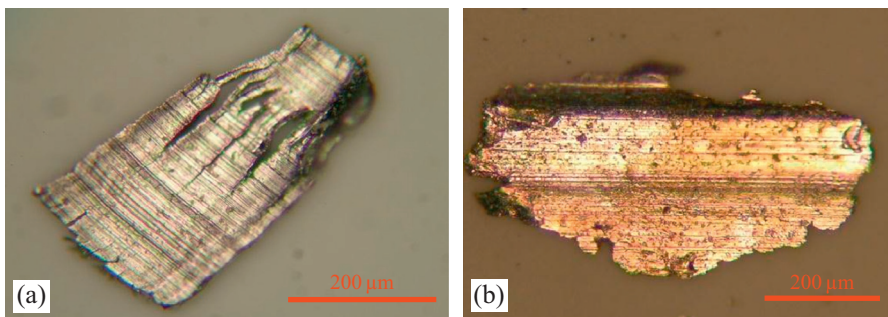


FIGURE 13.40

The large-size babbitt alloy wear particles from the oil drain of a hydro-turbine (200×): (a) a sliding babbitt alloy wear particle with the white metal appearance; (b) a babbitt alloy sliding wear particle with oxide film.

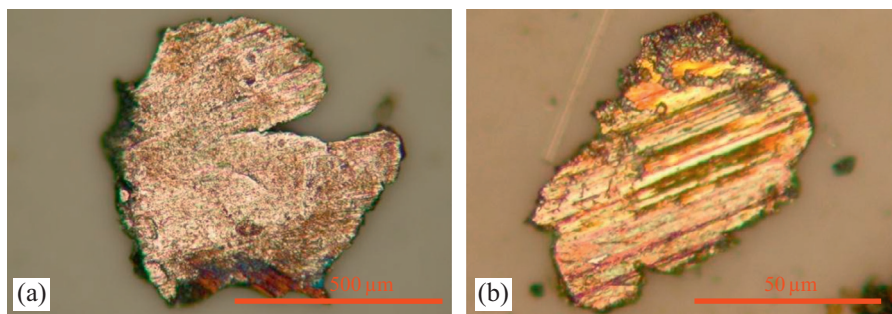


FIGURE 13.41

The babbitt alloy wear particles with oxide film (200 $\times$): (a) a babbitt alloy fatigue wear particle from an oil sample taken from the oil drain of a hydro-turbine; (b) a babbitt alloy sliding wear particle from a steam turbine-generator.

characterized by their uniform diameters and stiffness whereas cloth fibers feature cloth colors and the comparative uniform diameters. Generally, friction polymer fiber particles are characterized by their uneven diameters and the broken features (Figure 13.47b).

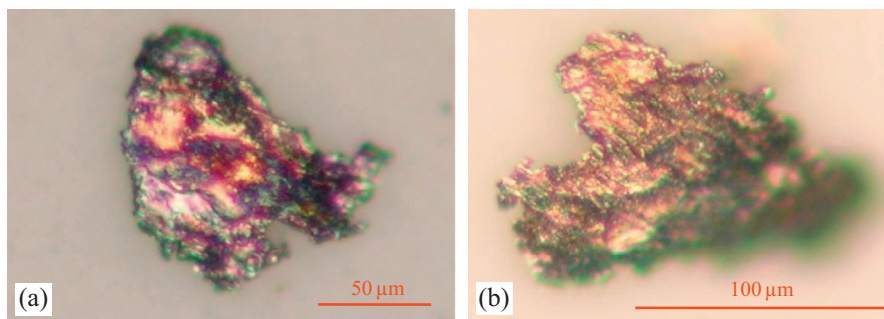
In the lubricating oil, the friction polymer particles can melt. Then the molten polymer particles can be drawn into fiber particles by mechanical movement. The evidence of the friction polymer fiber formation has been discovered on from the bearing oil samples (Figure 13.48).

12.2 VARNISH PARTICLES

Varnish deposit is a vicious lubricant failure as it not only causes the machine malfunctions [18–20] but also is difficult to diagnose and remedy [21–24]. When the varnish deposit occurs, significant amount of varnish particles would exist in the oil. Therefore, identifying varnish particles on the filtergram aid monitoring of varnish buildup.

Varnish particles appear in different colors, ranging from light-brown to dark-brown. The color variations associate with different stages of oil degradation [24,25]. The dark color varnish particle could be associated with some degree of aromatization reaction [25]. Figure 13.49 displays the different color varnish particles.

The compositions of varnish particles are usually especially complicated. Figure 13.50 shows the ESEM/EDS analysis on the above light-brown varnish particle (Figure 13.49a). As EDS cannot detect hydrogen, the high carbon and oxygen contents in the results indicate that the varnish was from the oil oxidation. Considerable amount of iron in the results also reveals that some submicron metal wear

**FIGURE 13.42**

The babbitt alloy wear particles with different oxide colors (200 $\times$): (a) a babbitt alloy fatigue wear particle resembles to copper alloy wear particle; (b) a babbitt alloy fatigue wear particle resembles to steel wear particle.

particles were embedded inside the varnish particle. Also trace amount of Zn, S, P, and Mg are present, indicating that the oil additives also involved the varnish formation.

Like lacquer, varnish deposit can be hardened. Hence, the hardened varnish particles can be generated. [Figure 13.51](#) shows the hardened varnish particles. [Figure 13.51a](#) is a hardened varnish particle with the rock features. This particle was from a bearing oil sample. [Figure 13.51b](#) is hardened varnish particle with the “mud-crack” features, showing the brittleness. This particle was from a gearbox oil sample. The brittle varnish particles can be fragmented into much smaller varnish particles.

Brown is the base color for the majority of varnish particles. However, various colors of the varnish particles can be occasionally found in the oil samples. [Figure 13.52](#) displays the different color varnish particles, ranging from green to purple. In [Figure 13.52b](#), some of the miniscule metal wear particles embedded inside the varnish particle can be identified.

12.3 ADDITIVE DEGRADATION PARTICLES

Generally, the additives can be completely dissolved in the lubricating oils. In some cases, some additives would precipitate from the oil, forming the solid particles or film debris. [Figure 13.53](#) shows the film debris precipitated from a bearing oil sample. Like the natural oil film on water, the film debris appeared in beautiful colors due to light diffraction. [Figure 13.54a](#) shows other film debris from the same bearing oil. [Figure 13.54b](#) was film debris from a gearbox oil sample.

The above film debris ([Figure 13.54a](#)) had been analyzed by ESEM/EDS. [Figure 13.55a](#) was the debris' ESEM image. *In situ* EDS analysis results on the debris were shown in [Figure 13.55b](#). The high carbon and oxygen contents in the results confirm the organic compositions. The substantial high zinc and significant phosphorous indicate that the substance was from the EP additives.

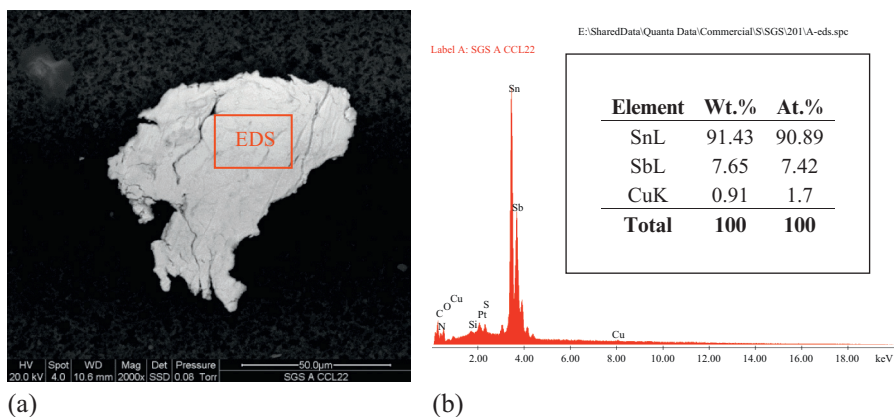


FIGURE 13.43

The ESEM/EDS analysis results on the particle (Figure 13.42a): (a) the particle's ESEM image (2000×); (b) the EDS analysis results.

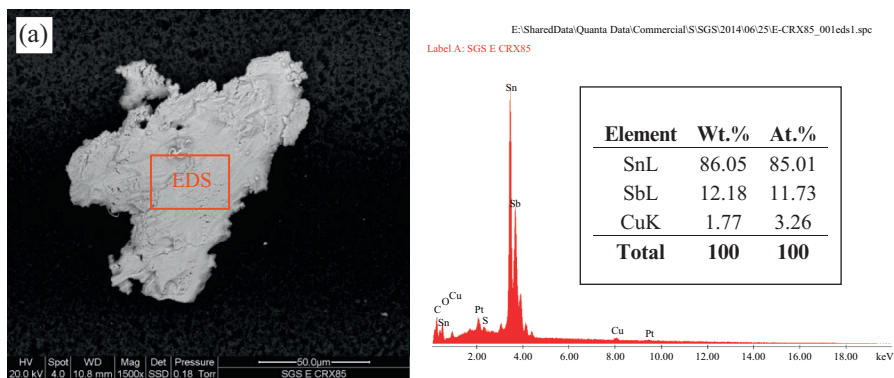


FIGURE 13.44

The ESEM/EDS analysis results on the particle (Figure 13.42b): (a) the particle's ESEM image (1500×); (b) the EDS analysis results.

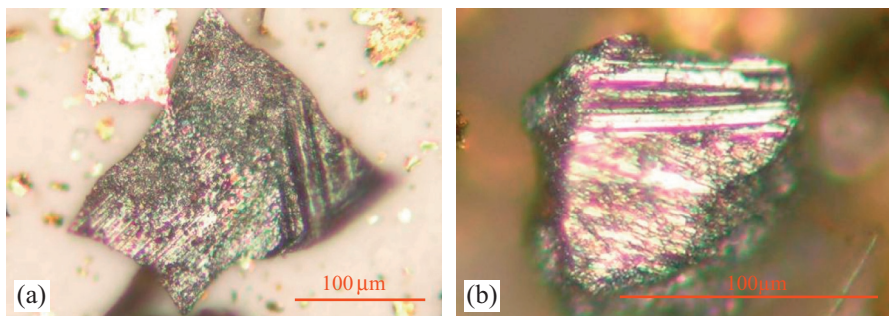
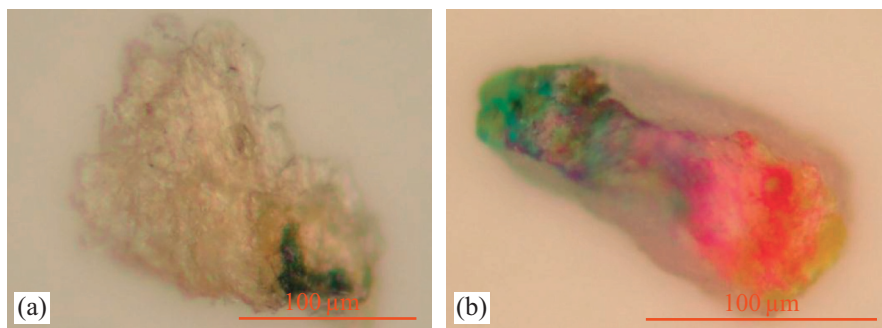
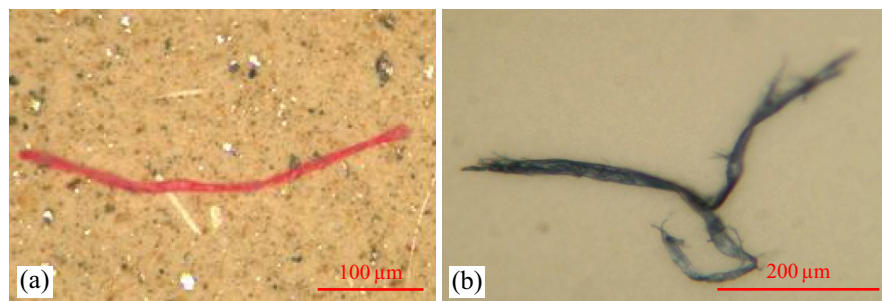


FIGURE 13.45

The typical babbitt alloy wear particles from an oil filter of a geothermal steam turbine-generator (200×): (a) a babbitt alloy fatigue wear particle; (b) a babbitt alloy fatigue wear particle.

**FIGURE 13.46**

The friction polymer particles ($200\times$): (a) a friction polymer particle from a new oil sample; (b) a color friction polymer particle from a gas turbine oil sample.

**FIGURE 13.47**

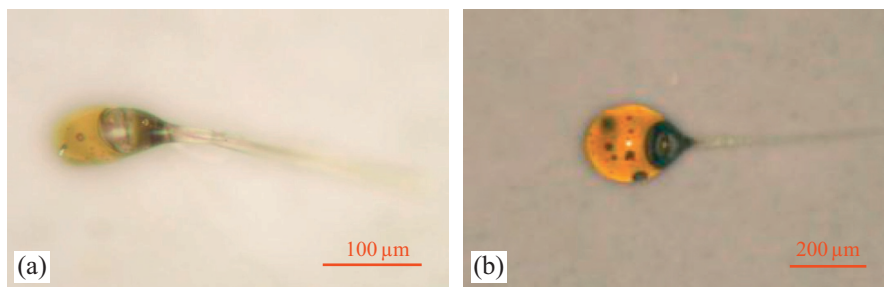
The friction polymer fiber particles ($200\times$): (a) a friction polymer fiber particle from a gearbox oil sample; (b) a friction polymer fiber particle from a gas turbine oil sample.

12.4 PSEUDO-METAL WEAR PARTICLES

Some varnish particles would appear to be shining due to their angles of orientation on the optical microscope. In this case, the varnish particles resemble to metal wear particles. This type of varnish particles is designated as pseudo-metal wear particles.

Figure 13.56a was a pseudo-metal wear particle, which resembles to a metal wear particle with dark oxide film. However, gentle touching by the tip of a needle can fragment this particle (Figure 13.56b). Actually, the particle was a brittle varnish in which some minute metal wear particles were embedded. Hence, the varnish particle would behave like steel wear particle under the magnetic field. However, this type of pseudo-metal wear particles should not exist in ferrogram because they would be fragmented by the micropump in the ferrograph instrument.

Figure 13.57 shows another type of pseudo-metal wear particle. This particle was from a gas turbine seal oil sample. At $100\times$ magnification (Figure 13.57a), the

**FIGURE 13.48**

The evidence of the formation of friction polymer fiber from the bearing oil samples: (a) the molten friction polymer bead and the fiber formation ($200\times$); (b) the molten friction polymer bead and the fiber formation ($50\times$).

shining feature resembles to the metal wear particles. However, at $200\times$ magnification (Figure 13.57b), the particle's surface indicates the typical attributes of varnish particles. Figure 13.58 displays other two similar pseudo-metal wear particles.

Figure 13.59 demonstrates other examples of pseudo-metal wear particles. These two particles were from wind turbine gearbox oils samples. The first particle (Figure 13.59a) resembles to a fatigue wear particle with partial dark oxide, whereas the second particle (Figure 13.59b) resembles to a severe sliding wear particle with partial dark oxide surface. Through ESEM/EDS analyses, both examples were confirmed to be varnish particles.

Figure 13.60a shows the secondary electronic image of one of the first particle. *In situ* EDS analysis results (Figure 13.60b) confirms the varnish compositions. Figure 13.61a displays the detailed features of the second particle. The deep striations on the particle surface indicate that the particle had been subjected to sliding wear. However, *in situ* EDS analysis results (Figure 13.61b) reveal its varnish compositions.

12.5 CASE STUDY 6: THE VARNISH DEPOSIT AND THE ONSET DAMAGE ON THE BEARING PADS OF A 55 MW GAS TURBINE LOAD GEARBOX

A load gearbox in a 55 Mw gas turbine experienced the varnish deposit issue. The gearbox was a one-stage reducing load gearbox which transmitted the power from the turbine to the generator. An oil tank (6.5 m^3) supplied the lubricant to all the babbit bearings of the turbine, generator, and the gearbox. The comprehensive turbine oil analysis program did not detect the abnormality, but filtergram analysis detected the varnish particles and the wear particles for some time. Figure 13.62 shows the typical varnish particles.

The varnish deposit on the bearing pads had inflicted damage to the bearing. Figure 13.63 shows the typical wear particles found in the oil samples. After the problems had been detected through the filtergram technique, the gearbox was overhauled

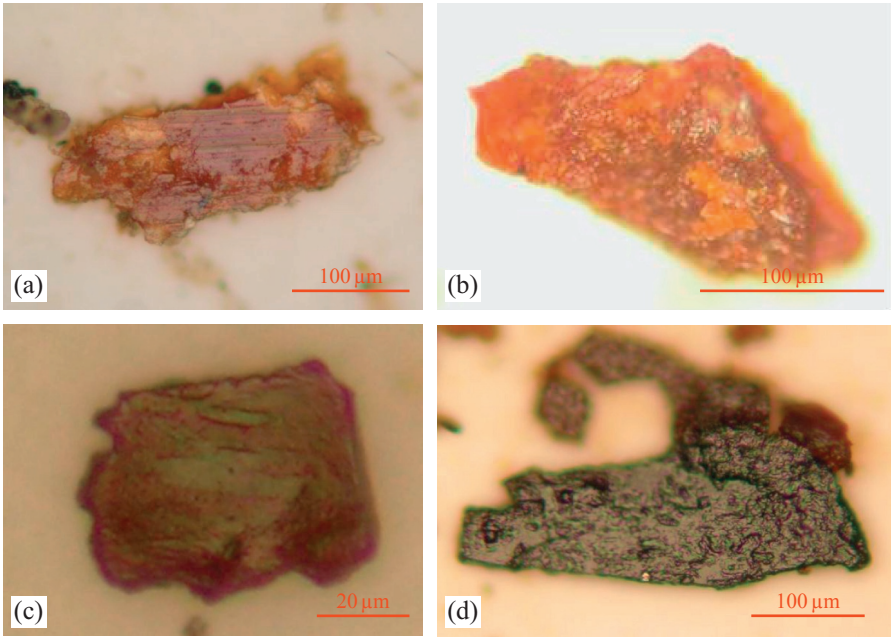


FIGURE 13.49
The different color varnish particles (200×): (a) a light-brown varnish particle from a compressor oil sample; (b) a brown varnish particle from a gas turbine oil sample; (c) a dark-brown varnish particle from a gearbox oil sample; (d) a dark varnish particle from a gearbox with varnish deposit issue.

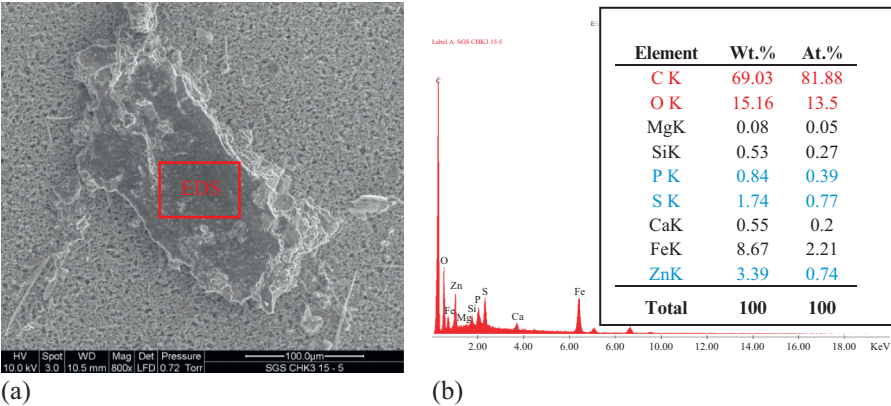


FIGURE 13.50
The ESEM/EDS analysis results on above light-brown varnish particle: (a) the particle ESEM image (800×); (b) the EDS analysis results.

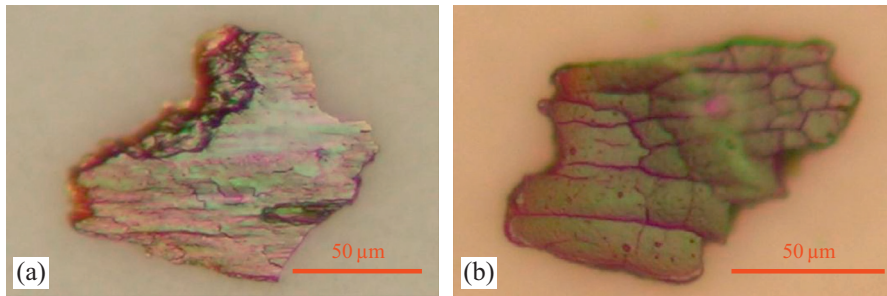


FIGURE 13.51

The hardened varnish particles (200 $\times$): (a) a varnish particle with the rock feature; (b) a varnish particle with the “mud-crack” features.

during shutdown. Varnish deposits on all the six babbit bearings were confirmed (Figure 13.64). The bearing pads also suffered different degrees of the damages.

In addition to the varnish particles and the babbit alloy wear particles, significant amount of rubber particles had also been found on the filtergrams. Figure 13.65a shows the typical rubber particles in this case study. In the overhaul, the root cause of the rubber particles was uncovered. As a piece of damaged rubber seal was clogged in the inlet of the gearbox bearing #5 (Figure 13.65b), significant amount of rubber particles had been generated in the oil.

13 THE MICROORGANISM DEBRIS

Some types of microorganisms can thrive at extreme conditions. Therefore, it is not surprising that microorganisms can be found in oil samples. Figure 13.66 show the microorganisms found in the new oil samples. Microorganisms can be easily noticed by their biopatterns.

In used oil samples, the debris of microorganisms can occasionally be found. Due to the machine running, the microorganisms could be shattered into debris. Figure 13.67 shows the examples of the microorganism debris from different used oil samples.

14 CONCLUSION

Wear particle analysis is a powerful technique in condition monitoring and wear diagnosis. Unlike ferrogram, filtergram can collect all types of particles, including varnish particles. Filtergram can reveal the particle’s detail features, which is much more accurate than computer image analysis.

In this chapter, a concise filtergram wear particle atlas is presented with the case studies. The six case studies are summarized in Table 13.1.

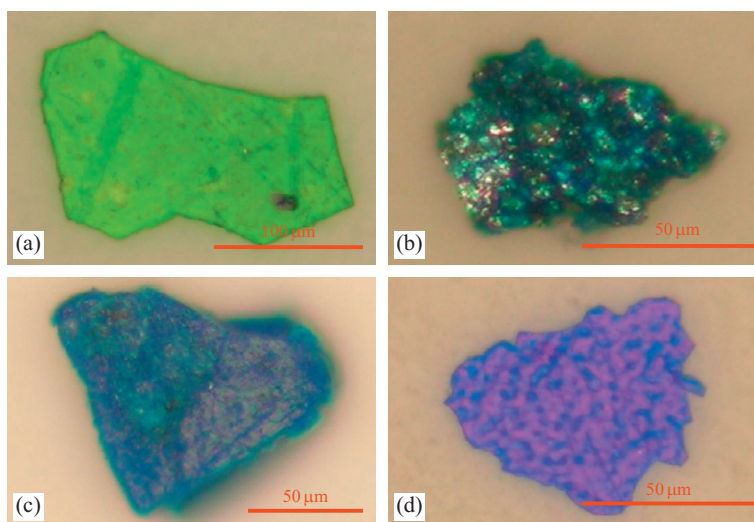


FIGURE 13.52

The different color varnish particles (200 ×): (a) a green varnish particle from a gas turbine oil sample; (b) a dark-green varnish particle from a hydraulic oil sample; (c) a blue varnish particle from a hydraulic oil sample; (d) a purple varnish particle from a gearbox oil sample.

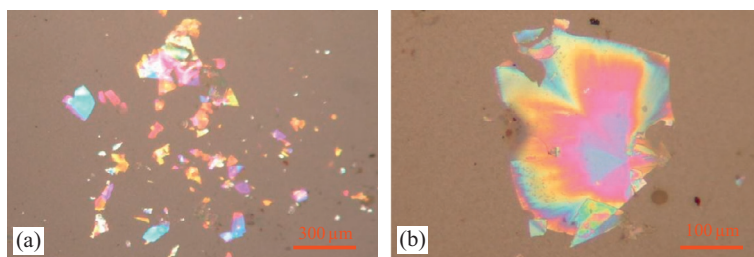


FIGURE 13.53

The beautiful colors of the film debris of the EP additives from a bearing oil sample (Mobil Vaculine 528) under optical microscope with white reflect light: (a) 50 ×; (b) 100 ×.

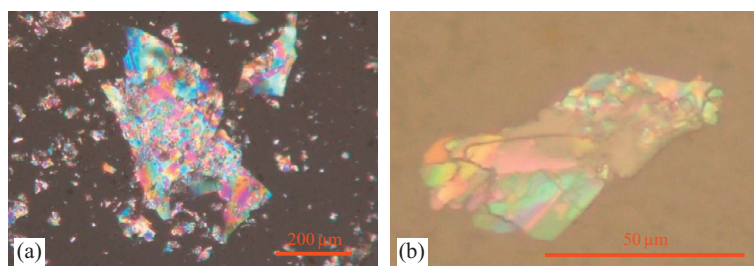


FIGURE 13.54

The film debris from different oil samples: (a) the film debris from a bearing oil sample (50 ×); (b) the film debris from a gearbox oil sample (100 ×).

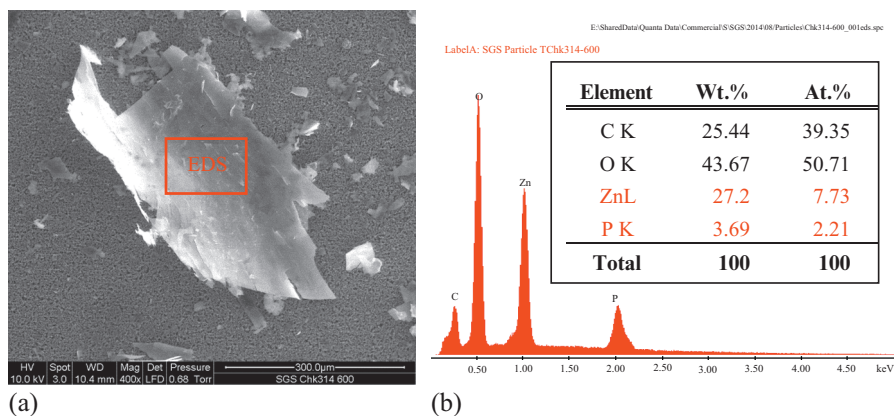


FIGURE 13.55
The ESEM/EDS analysis results on thin film debris: (a) the ESEM image of the thin film debris (400 ×); (b) the EDS analysis results.

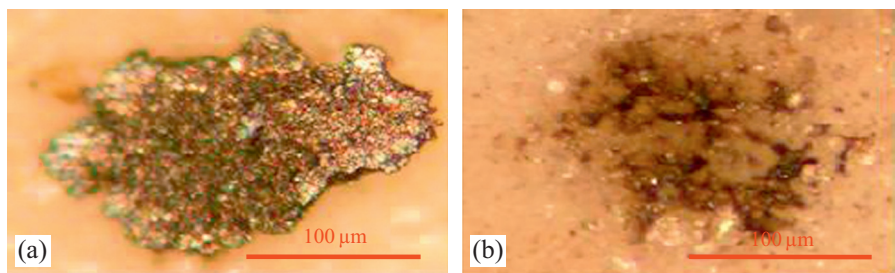


FIGURE 13.56
A pseudo-metal wear particle (200 ×): (a) reflective features of a pseudo-metal wear particle; (b) the pseudo-metal wear particle fragments into much smaller particles by a gently touch with a needle tip.

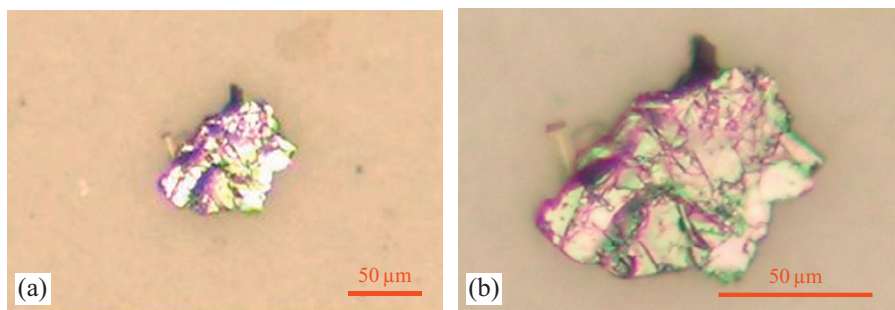


FIGURE 13.57
A pseudo-metal wear particle from the seal oil sample from a gas turbine: (a) a pseudo-metal wear particle at magnification of 100 ×; (b) the same particle at 200 × magnification.

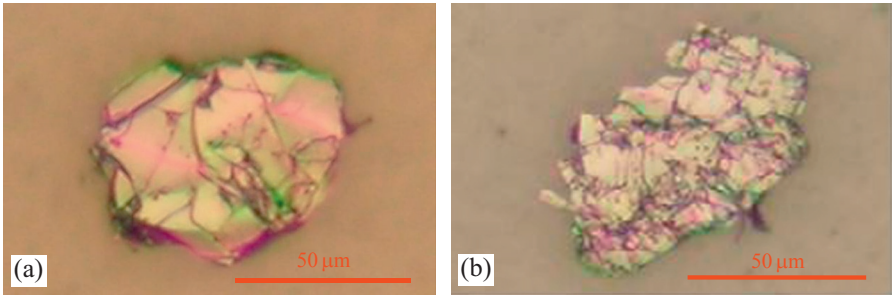


FIGURE 13.58
The pseudo-metal particles (200×): (a) a varnish particle from the seal oil sample from a gas turbine; (b) a varnish particle from a journal bearing oil sample.

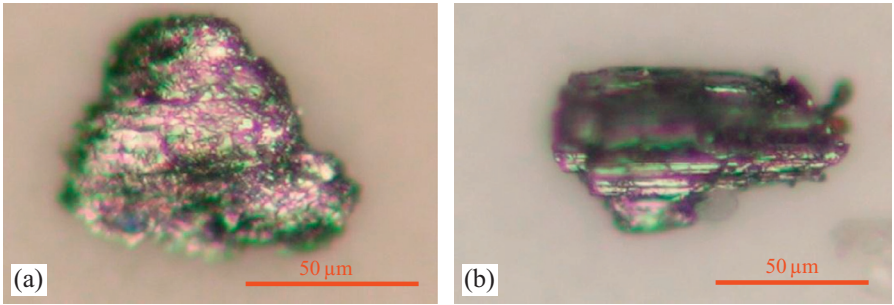


FIGURE 13.59
The pseudo-metal wear particles from the oil sample of a wind turbine gearbox (200×): (a) a pseudo-metal wear particle; (b) a pseudo-metal wear particle.

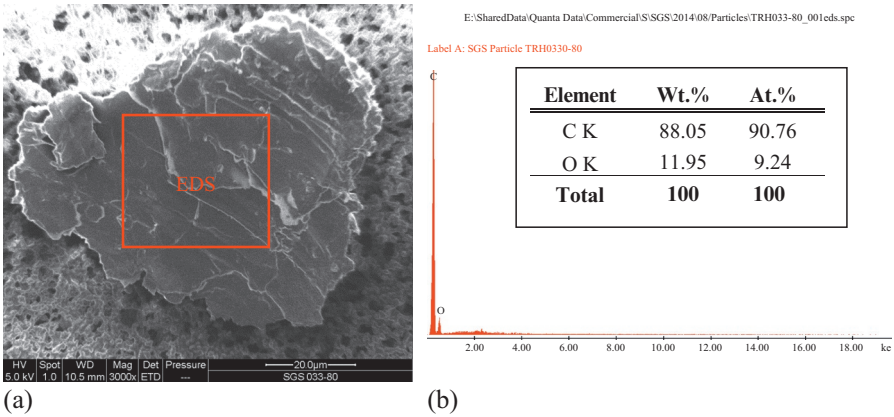


FIGURE 13.60
The ESEM/EDS analysis results on the pseudo-metal wear particle (Figure 13.59a): (a) the ESEM image of the pseudo-metal wear particle (3000×); (b) the EDS analysis results.

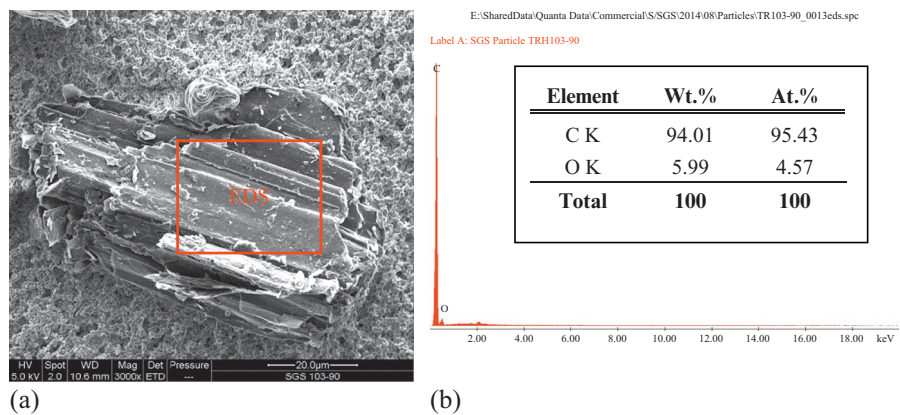


FIGURE 13.61

The ESEM/EDS analysis results on the pseudo-metal wear particle (Figure 13.59b): (a) the ESEM image of the pseudo-metal wear particle (3000×); (b) the EDS analysis results.

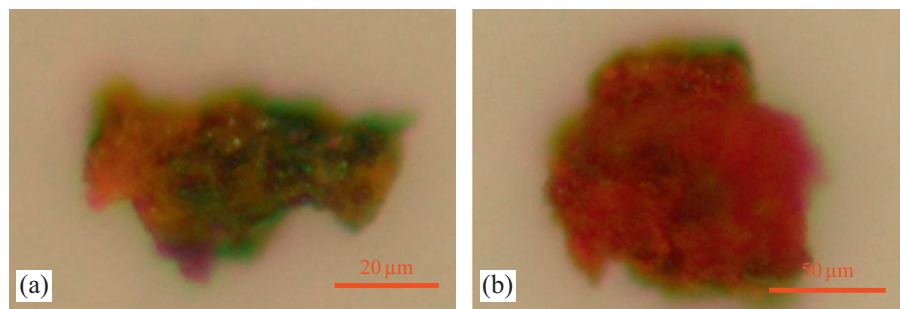


FIGURE 13.62

The varnish particles in the oil samples taken from the gas turbine oil tank (200×): (a) a varnish particle; (b) a varnish particle.

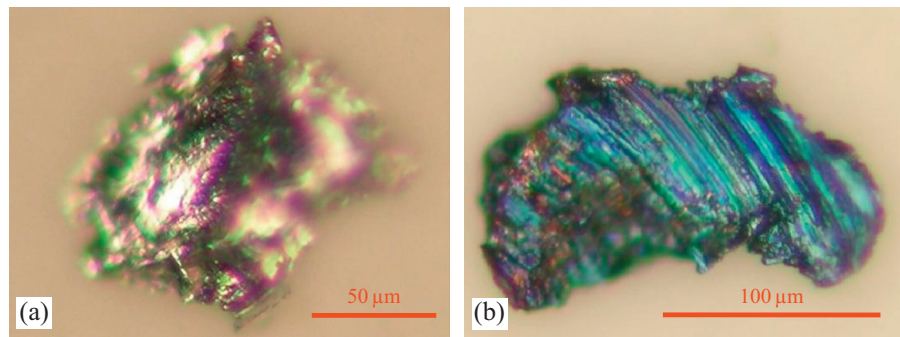


FIGURE 13.63

The severe wear particles (200×): (a) a babbit alloy wear particle from the bearing pads; (b) a severe sliding wear particle from the gears.

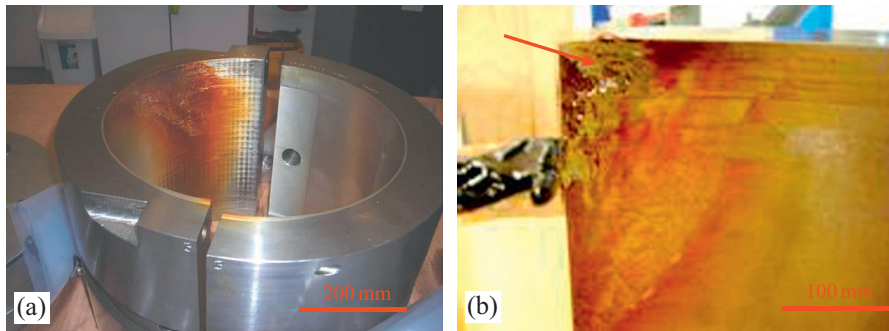


FIGURE 13.64

The varnish deposit and the damages on the gearbox bearing pads: (a) the varnish deposit and the onset damage on bearing #3; (b) the moderate damage in the edge of bearing #5 due to the varnish deposit.

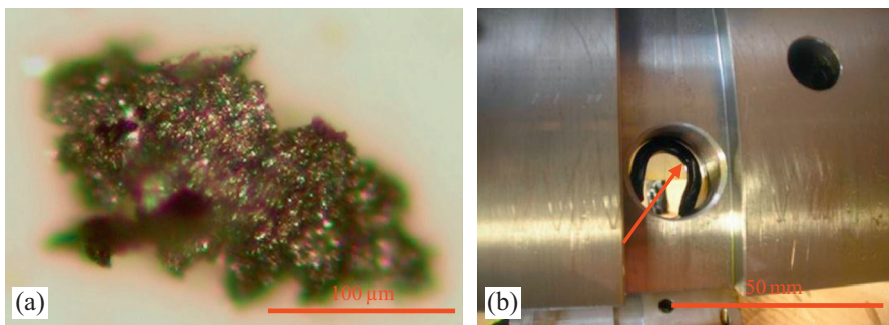


FIGURE 13.65

The damaged rubber seal generated significant amount of rubber particles: (a) a typical rubber particles on filtergram (200×); (b) the damaged rubber seal clogged in the inlet of bearing #5.

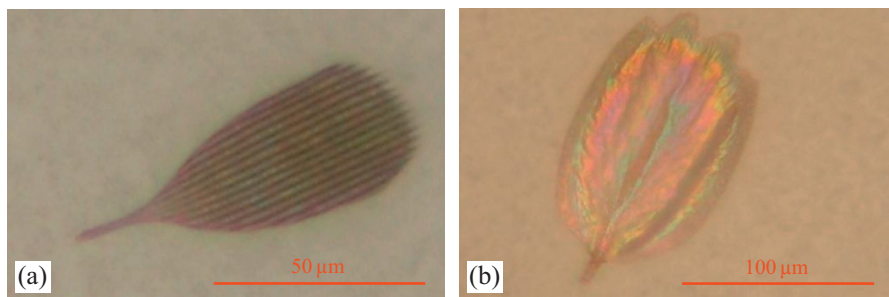


FIGURE 13.66

The microorganism from a new oil sample (200×): (a) a microorganism; (b) a microorganism.

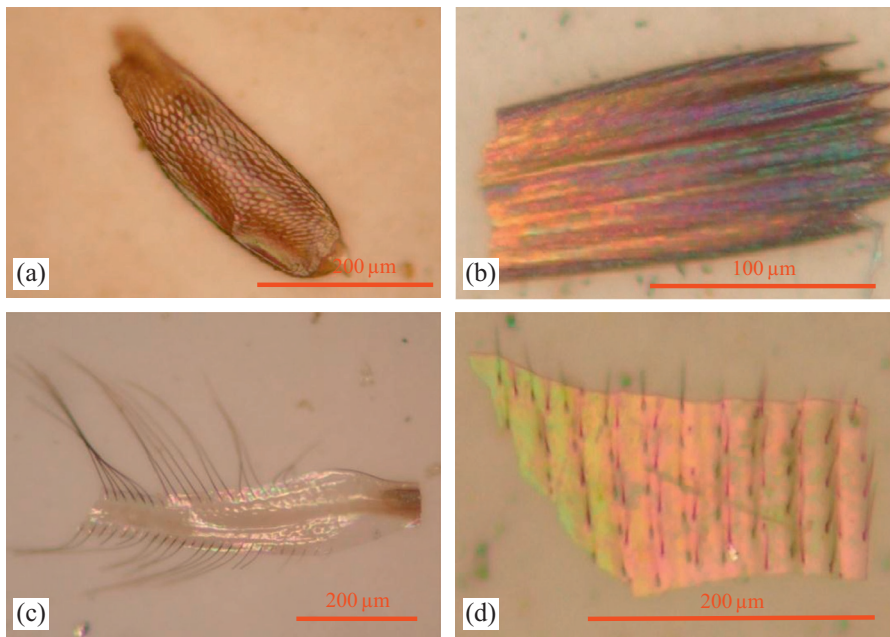


FIGURE 13.67

The microorganism debris from the used oil samples: (a) a microorganism debris from a gearbox oil sample (200 $\times$); (b) a microorganism debris from a bearing oil sample (200 $\times$); (c) a microorganism debris from geothermal steam turbine oil sample (100 $\times$); (d) a microorganism debris from a hydro-turbine oil sample (200 $\times$).

Table 13.1 Summary of the Case Studies

Case Studies	Machine Type	Failure Mode and the Root Cause of the Failure	Typical Wear Particles
1	Race car gearbox	Catastrophic wear due to inadequate lubrication (wrong type of oil)	Large-size wear particles with color oxide film
2	Wind Turbine gearbox	Bearing spalling by the white etching cracks (WEC)	Large-size wear particles (laminar particle, sliding wear particles, fatigue wear particle, and spherical particle)
3	Bulldozer diesel engine	Seizure failure by catastrophic wear due to inadequate lubrication	Large-size sliding wear particles, fatigue wear particles, and copper alloy wear particles
4	Yankee machine in a paper mill	Bearing failure due to high level of moisture content in the oil	Large-size fatigue wear particles with multi-oxide scales
5	250 Kw gearbox in a paper mill	Impending catastrophic wear due to the blockage of the oil supply	Large-size sliding wear particles and black oxide fatigue wear particles
6	55 Mw gas turbine load gearbox	The onset severe wear on babbit bearing pad due to varnish deposit	Vanish particles, babbit alloy wear particles, and sliding wear particles with blue oxide color

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Fatigue failure analysis of welded structures: An overview and A case study

14

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1 INTRODUCTION

Fatigue is a mode of failure and is responsible for most of failures in mechanical components and structures [1]. Fatigue is the process of progressive and localized structural change in a material under a fluctuating loading [2]. In this definition, the term “progressive” means the fatigue process occurs over a period of time, although the fatigue failure happens abruptly. The term “localized” states that the fatigue process operates at local regions rather than throughout the whole component

or structure, and the word “fluctuating” means that the loading associated with fatigue is alternating rather than being stationary [1].

Although the number of mechanical failures cases is low when compared to successful components and structures, the cost associated with the failures is overwhelming. A comprehensive study was carried out in 1982 to estimate the cost of fractures in the United States. This study shows that the restitution of mechanical fractures in 1 year was \$119 billion, which could have been decreased significantly using appropriate engineering designs, including fatigue design [3].

Fatigue failure in metals involves three stages: fatigue crack initiation, crack propagation, and final fracture, Figure 14.1. When a metallic component is loaded, shear stress may develop parallel planes which tend to move relative to each other. *Slip* plastic deformation happens when a plane slides against adjacent planes within a grain. As they are the weakest, surface grains are more prone to slip plastic deformation, which produces a microcrack [4].

Fatigue cracks in many cases initiate from the surface microcracks, to the length of tenth of millimeters [5]. Microcracks usually initiate on the plane of maximum shear stress [1]. For the case of uniaxial loading, maximum shear stress planes make an angle of 45° with the loading direction, and slip takes place at about this angle, Figure 14.1 (stage I).

In some cases, small fatigue cracks join together and propagate often leaving dark and light bands on the fracture surface. Fatigue crack propagation involves plastic deformation in small regions around the crack tip, but global plasticity and necking is usually minimal; therefore, fatigue failure is often known as brittle failure. Fatigue

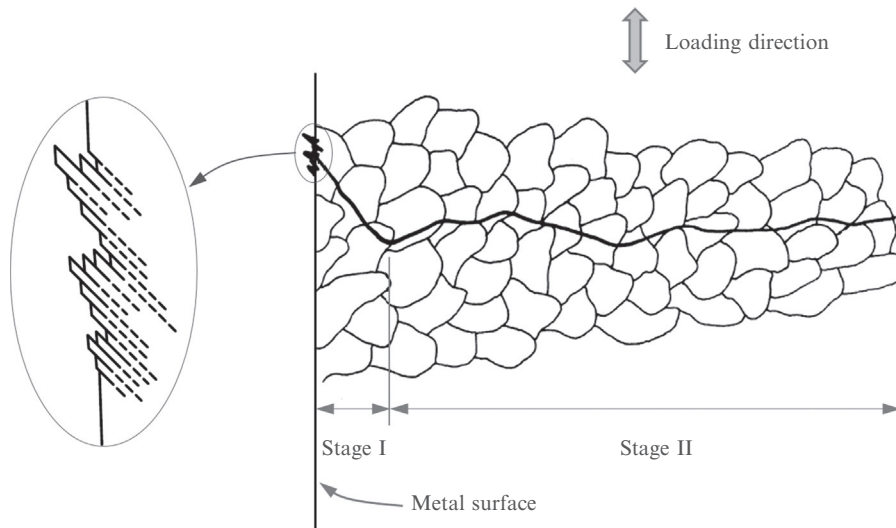


FIGURE 14.1

Schematic of fatigue failure stage I (crack initiation) and stage II (crack propagation) [1].

cracks often propagate on the plane of maximum normal stress (with exceptions of torsional and multiaxial loading) [1]. When the fatigue crack has reached a critical length, final fracture will occur. Fatigue cracks prior to fracture may range from very small cracks (1% of the fracture surface) to very large cracks ($\sim 100\%$), depending on the load level and fracture toughness of the material [1].

The research on the fatigue of metals started in the nineteenth century [5,6], and since then numerous studies have been performed on this topic. These studies followed two major approaches: fatigue characterization and fatigue modeling. The fatigue characterization approach investigates the macro- and microstructure of materials before, during, and after cyclic loading and explains the mechanism of fatigue failure. The fatigue modeling approach predicts the fatigue failure location and service life of a component or structure under a given loading condition. With all the studies, this phenomenon is still not fully understood because of the complexities involved in the fatigue process [4,7], such as micromechanism of cyclic plastic deformation, and determination of the crack propagation path.

Fatigue of welds is even more complex. Welding processes often cause liquation and rapid solidification which result in different microstructure in the weld region than in the base metal. Deposition of filler metals in arc welding increases the inhomogeneity of the weld region. Residual stresses induced in the weld region due to the solidification and shrinkage affect the fatigue strength. Also, stress concentration around the weld and imperfections inside the weld, for example, porosity, undercuts, etc., add to the complexity of the fatigue process [8].

In welded structures, the weld regions are very prone to fatigue failure because of the stress concentration, tensile residual stress, etc. Therefore, assessing the fatigue strength of welds has a great significance. This topic has attracted researchers' attention and several models have been proposed to predict the fatigue life of welded structures. From the fatigue modeling perspective, welded joints are often divided into seam welds and spot-welds [8,9], because they are quite different in stress analyses. The focus in this chapter is on spot-welds; therefore, the common approaches for fatigue modeling of spot-welds are briefly introduced. It is noteworthy that the fatigue models explained in this chapter have been developed based on the fatigue test results of resistance spot-welded specimens, Figure 14.2; however, their

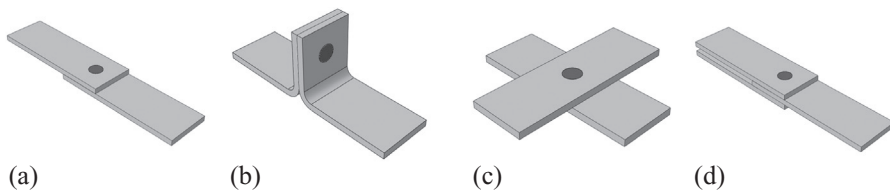


FIGURE 14.2

Typical spot-welded specimens: (a) tensile-shear, (b) coach-peel, (c) cross-tension, (d) double-shear [10].

application can be extended to any spot joint, such as friction stir spot-welds and self-piercing rivets, due to the similarity of the geometry and stress analyses as well as flexibility of the fatigue models.

Fatigue modeling of weldments, in general, and spot-welds, in particular, is based on the comparison of loads, stresses, or strains with their critical or strength values which cause a defined damage or total fracture [9]. These models can be categorized into global and local approaches. The global approaches assess the fatigue strength using the external forces and moments, or using the nominal stress at the critical cross-section calculated from the external forces. Fatigue life is determined by comparing this value with the corresponding critical value related to a global failure such as fully yielding or total rupture. The local approaches, in contrast, use local parameters at the critical point to evaluate the fatigue damage. Local approaches are capable to follow the fatigue crack initiation and crack propagation processes. To predict the crack initiation behavior, local notch approach is applied, that is, a measure of local notch stress or strain is utilized as the damage parameter. Crack propagation behavior may be modeled by fracture mechanics approach in which stress intensity factor (SIF) or J -integral is the fatigue parameter. The transition between the global and local approaches is called the structural stress approach, which is sometimes considered as a local approach [9]. The structural stress is calculated from the internal forces at the weld (as opposed to external forces in the global approach), but does not explicitly account for the weld notch stress concentration.

Up until the 1970s, the global approaches were being used for fatigue assessment of spot-welds, that is, the fatigue strength was represented by specimen-dependent quantities, such as the load amplitude. These approaches required an extensive library of fatigue strength data for different spot-weld configurations and loading conditions. Also, the global approaches are not capable of following the fatigue process which has a local nature. These drawbacks have limited the application of global approaches, although they are very simple to apply. Local approaches on the other hand have attracted attentions since then, and have been widely applied to spot-welds.

The fracture mechanics approach considers the spot-weld as a sharp notch, and treats it as a crack [11–13]; therefore, the crack initiation life is often assumed to be insignificant in this approach. The fatigue models based on the fracture mechanics use a measure of SIF or J -integral as the fatigue damage parameter and relate this parameter to the fatigue life or the crack growth rate. Pook in 1975 [14] derived formulations for the modes I and II SIFs for spot-welds under tensile-shear loading, Figure 14.2a, and the range of mode I SIF, ΔK_I , was introduced as a fatigue damage parameter. Swellam in 1992 [15] developed a fatigue model for spot-welds, combining the effects of mode I and mode II SIFs. This model considered spot-welds under a general load case. Forces and moments at the spot-weld center were found from static equilibrium equations, and SIFs were calculated according to the work by Tada et al. [16]. Assuming that the SIFs are independent of the crack length, fatigue life was calculated from the SIFs through the well-known Paris' equation. Newman in 1998 [17] developed a fatigue model for spot-welds under the tensile-shear load case. Formulations proposed by Pook [14] for mode I and II SIFs were employed and combined to obtain an

equivalent mode I SIF on the plane of maximum principal stress. Finally, a closed form equation was derived between the equivalent model I SIF and the fatigue life considering a number of simplifications.

The structural stress approach has become popular since the late 1980s. Structural stress is a linearly distributed stress over the thickness, obtained by neglecting the effect of stress concentration arisen from the weld. The structural stress is usually calculated by superposing the effects of different forces and moments, which are obtained from linear elastic finite element (FE) simulations. FE modeling of spot-welded structures in this approach is performed such that the sheets and spot-welds are modeled by shell and beam elements, respectively. The structural stress approach, due to its flexibility, is extensively employed in industries [9]. The models proposed by Radaj [18], Rupp [19], and Sheppard [20] are among the most frequently used models based on the structural stress approach. Radaj in 1989 [18] proposed the concept of structural stress. Assuming that the spot-weld is an ideal crack-like slit, SIFs for mode I, II, and III load cases were related to nominal structural stresses. SIFs were combined to obtain an equivalent SIF, which was used for fatigue life estimation. Radaj in 1990 [21] presented formulations for structural stresses in terms of forces and moments. Rupp et al. in 1995 [19] employed theories of beams and shells to develop a structural stress method for spot-welds. Forces and moments were obtained at the nugget center from a FE analysis. Through-nugget cracking and through-plate cracking were considered as possible failure modes in spot-welded structures. The beam theory was used to find structural stresses for through-nugget failure, and the theory of shells was employed to obtain structural stresses for through-plate failure. Sheppard et al. in 1992 [22] developed a model to predict fatigue crack initiation life for spot-welds based on the structural stress concept. The structural stress in this model was calculated from the membrane load and bending moments. Structural stress range was obtained on the shell elements neighboring the spot-weld at the node shared between the shell element and the spot-weld beam element. The maximum structural stress range was converted to actual elastic-plastic stress and strain ranges using the Neuber's rule [23]. The Smith-Watson-Topper relationship [24] was employed to find fatigue initiation life in terms of the actual stress and strain ranges. Sheppard in 1993 [25] extended the application of the structural stress to estimate the fatigue crack propagation life.

The local notch stress/strain approach, in contrast to the fracture mechanics approach, considers a spot-weld as a blunt notch with a finite radius. Therefore, a detailed FE model with a fine mesh in the vicinity of the spot-weld is required. The fatigue life in these models is usually related to a measure of local stress or strain at the spot-weld edge. Local stress/strain values are calculated from an elastic-plastic FE simulation, or from an elastic solution along with a variant of the Neuber's rule. The most commonly referenced models in this group were developed by Oh [26] and Pan [27]. Oh in 1982 [26] obtained stress/strain results from linear elastic FE analysis. The elastic solution in the region around the spot-weld nugget was transferred to an elastic-plastic solution using the Neuber's rule, and the estimated stress and strain values were related to total fatigue life through the Morrow's rule [28]. Pan in 2000 [29] demonstrated that Sheppard's structural stress model was not successful for

spot-welds with dissimilar thicknesses. Therefore, Pan [27,29] proposed a local strain model for such cases. Local stress and strain values at the spot-weld edge were obtained from a three-dimensional elastic-plastic FE simulation. Nonlinear geometry and nonlinear material properties were included in the FE analysis. Cyclic principal strain range was considered as the fatigue damage parameter and was related to total fatigue life.

This chapter briefly explains the procedure for fatigue failure analysis of weldments, with a special focus on spot-welded structures. The components involved in the fatigue modeling are introduced and described. Different approaches for fatigue modeling of welded structures are presented and their advantages and drawbacks are discussed. A real-life welded structure, which is essentially a subassembly of automotive body parts, is studied for fatigue strength analysis. The fatigue modeling is performed using a global and a local approach. The prediction results, in terms of the fatigue failure location and the fatigue life, are compared to the experimental results and the results are discussed.

2 FATIGUE MODELING OF WELDED STRUCTURES BY LOCAL APPROACHES

Several fatigue models based on the local approaches have been proposed for spot-welded structures in the past four decades. These models can be categorized into three major groups: fracture mechanics, structural stress, and local notch stress/strain approaches. General features of these approaches are explained in this section, and a well-known fatigue model from each approach is explained.

2.1 FRACTURE-MECHANICS APPROACH

The fracture mechanics approach was one of the first solutions to avoid the problems associated with the global approaches. This approach has been applied to spot-welds since the 1970s. The spot-welding process produces a circular joint between two or more sheets, with a notch at the spot-weld edge. Because the notch radius is small compared to the sheet dimensions, the spot-weld in some studies is considered a sharp notch [11–13,30]. Therefore, the spot-weld is treated as a circular region surrounded by a pre-existing crack.

Furthermore, dependency of the fatigue strength to the material's static properties may help to identify an appropriate fatigue modeling approach. For the case of smooth specimens and blunt notched components, where the crack initiation process dominates, fatigue life is closely related to the material strength [31]. However, fatigue life in sharp notched and cracked components is controlled by crack propagation and is insensitive to material strength [32]. Experimental results for steel spot-welds demonstrate that the base metal strength has an insignificant effect on the fatigue strength [13,31]. These observations support the idea that spot-welds are crack-like flaws, and fracture mechanics is the right approach.

The fracture mechanics approach considers a measure of SIF or J -integral as the fatigue damage parameter, and relates this parameter to the fatigue life or the crack-growth rate. Pook [14], Swellam et al. [15], and Wang et al. [33] proposed different methods for calculating the SIF or J -integral range for spot-welds.

Swellam in 1992 [15] proposed a fatigue model for predicting crack propagation life for spot-welds based on the fracture mechanics approach. In this model, the effects of modes I and II loading were taken into account. The spot-weld was considered under a general applied load, F , as shown in Figure 14.3.

Resultant forces and moments, including axial load, P , shear load, Q , and bending moment, M , at the spot-weld center, were found from static equilibrium of a coupon. The SIFs developed by Tada et al. [16] for two half spaces joined by a circular region were used in this study,

$$K_I = K_{\text{axial}} + K_{\text{moment}} = \frac{P}{2r\sqrt{\pi r}} + \frac{3M}{2r^2\sqrt{\pi r}}, \quad (14.1)$$

$$K_{II} = K_{\text{shear}} = \frac{Q}{2r\sqrt{\pi r}}, \quad (14.2)$$

where r is the nugget radius. The equivalent mode I SIF, K_{Ieq} , was defined as

$$K_{Ieq} = \sqrt{K_I^2 + \beta K_{II}^2}, \quad (14.3)$$

in which β is an empirical material constant which reflects the material's sensitivity to mode II loading. The parameter β is obtained from experimental results of two or more different spot-weld specimen configurations. Some of the specimens must involve only the mode I SIF (such as cross-tension and coach-peel configurations, Figure 14.2), and some other specimens must reflect only the effect of mode II or

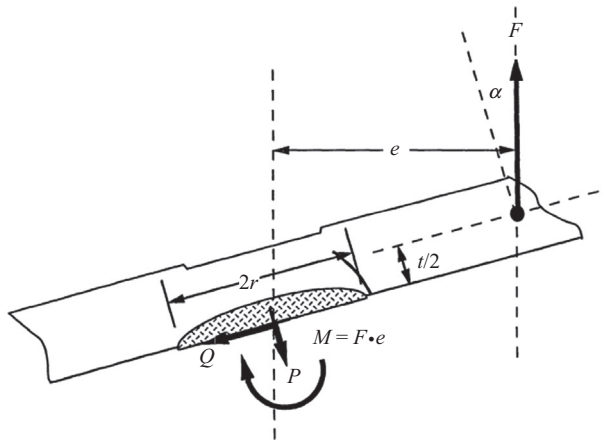


FIGURE 14.3

Resolving a general applied load, F , at the center of the spot-weld [15].

a combination of modes I and II SIFs (such as double-shear and tensile-shear configuration, Figure 14.2). The parameter β should be found such that the best correlation is achieved for equivalent stress intensities, K_{Ieq} , when plotted versus the fatigue life for all specimen configurations.

A geometrical correction factor, G , was defined to incorporate the effects of specimen and nugget size

$$G = \sqrt{\frac{t^2 W}{r^3} \left(\frac{9t^2}{4r^2} + 1 \right)}, \quad (14.4)$$

where t is the sheet thickness, and W is the specimen width. To account for the effect of load ratio, R , the general stress intensity parameter, K_i , was defined as

$$K_i = \frac{K_{Ieq}}{G} \sqrt{1 - R}. \quad (14.5)$$

The parameter K_i was related to the fatigue life by a linear relationship in the log-log scale.

The fracture mechanics approach in general has a number of advantages and drawbacks. The main advantage of this approach is that the crack propagation process may be closely followed. A drawback of this approach is that the nugget edge is considered a crack and therefore crack initiation life is assumed insignificant, which is not supported by experimental observations and analysis. The work by Swellam et al. [15] shows that crack initiation life in spot-welds of a low-carbon steel overwhelms the crack propagation life in the high-cycle fatigue regime, that is, life $> 10^5$ cycles. McMahon et al. [34] demonstrated that up to 55% of the total fatigue life of spot-welds can be consumed by crack initiation. This ratio in the work by Sheppard et al. [22] is 30%. It should be mentioned that the crack initiation life is the number of cycles required for the crack to reach a length of 18% of the sheet thickness in Swellam's work [15], and 0.25 mm in the studies by McMahon et al. and Sheppard et al. [22,34]. Another deficiency associated with the fracture mechanics approach is that the formulations are based on the assumption that the crack is along the faying surface, while according to experiments, through-thickness cracking is the most common fatigue failure mode for tensile-shear specimens [13,35–37]. Moreover, in contrast to the assumption in this approach, primary cracks in some cases do not initiate from the nugget edge [29,37].

2.2 STRUCTURAL STRESS APPROACH

Since 1989, a number of fatigue models have been developed for spot-welds based on the structural stress concept. The structural stress takes into account all the stress concentrations, except the stress concentrations originating from the weld [9,38]. Structural stress reflects the effects of forces and moments at the spot-weld center or edge, and theoretically may be defined in the plate or nugget depending on the mode of failure. Therefore, structural stress has been considered as the factor which controls the fatigue failure. The structural stress approach, as opposed to many

fracture mechanics-based models, often provides enough flexibility to be applicable to different specimens and structures. Therefore, these models are widely employed in industry, including the automobile industry [9]. However, some weak points are also associated with this approach. A serious shortcoming is that the notch effect at the spot-weld edge is ignored [9].

Sheppard and Strange in 1992 [22] studied steel spot-welds under cyclic loading and observed that failure occurs in the through-thickness fashion. They defined the structural stress as a function of membrane load and bending moment,

$$\Delta S_{ij} = \frac{\Delta P_{ij}}{\omega t_i} + 6 \frac{\Delta M_{ij}}{t_i^2 W}; \quad i, j = 1, 2, \quad (14.6)$$

where ΔS_{ij} is the structural stress range, ΔP_{ij} and ΔM_{ij} are membrane load and bending moment ranges, respectively. ω is the effective sheet width and according to Wang et al. [39] was defined as $\pi d/3$, in which d is the nugget diameter. The subscript i represents the sheet ($i=1$ for the bottom sheet, and $i=2$ for the top sheet), and the subscript j represents the side ($j=1$ for the right side, and $j=2$ for the left side). Figure 14.4 illustrates the membrane forces and bending moments at the edges of a spot-weld, that is, D_{ij} .

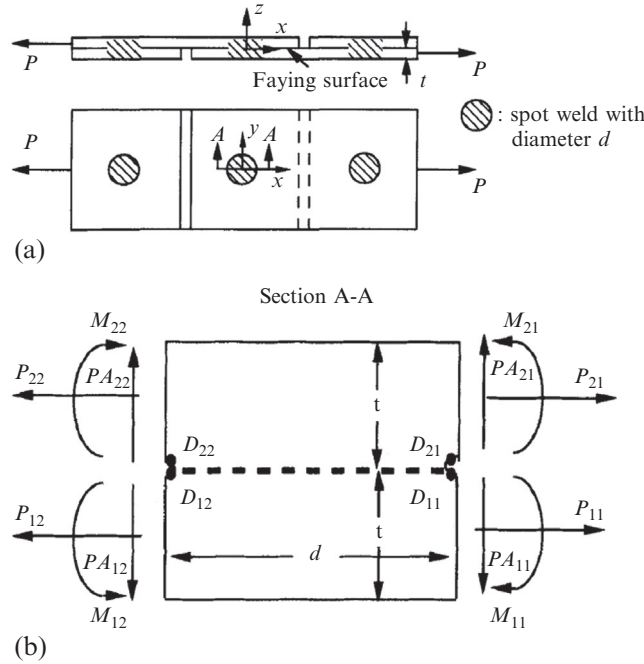


FIGURE 14.4

Forces and moments at a spot-weld nugget edge for Sheppard's structural stress calculation [22].

Forces and moments were determined from linear elastic FE simulations. As mentioned earlier, in the structural stress approach, sheets and spot-welds are modeled with shell and beam elements, respectively.

According to Equation 14.6, four values were obtained for the structural stress range at each spot-weld. Maximum structural stress range,

$$\Delta S_{\max} = \max(\Delta S_{11}, \Delta S_{12}, \Delta S_{21}, \Delta S_{22}), \quad (14.7)$$

was used in the Neuber's rule to estimate the maximum local stress and strain ranges, that is, $\Delta \sigma_{\max}$ and $\Delta \epsilon_{\max}$ (assuming cyclic Ramberg-Osgood behavior),

$$\frac{(K_f \Delta S_{\max})^2}{E} = \Delta \sigma_{\max} \Delta \epsilon_{\max}, \quad (14.8)$$

$$\Delta \epsilon_{\max} = \Delta \sigma_{\max} / E + 2(\Delta \sigma_{\max} / 2K')^{1/n'}, \quad (14.9)$$

where K_f is the fatigue notch factor, E is the elastic modulus, K' is the cyclic strength coefficient, and n' is the cyclic strain hardening exponent. The effect of mean stress, σ_m , on the crack initiation process was accounted for in this model and was obtained from

$$\sigma_m = \sigma_{\max} - 0.5 \Delta \sigma_{\max}, \quad (14.10)$$

in which $\Delta \sigma_{\max}$ was determined by solving Equations 14.8 and 14.9, and $\sigma_{\max}$ was found from the following equations

$$(K_f [\max \{S_{ij\text{peak}}\}] + \sigma_{rs})^2 / E = \sigma_{\max} \epsilon_{\max}, \quad (14.11)$$

$$\epsilon_{\max} = \sigma_{\max} / E + (\sigma_{\max} / K)^{1/n}, \quad (14.12)$$

where $S_{ij\text{peak}} = \Delta S_{ij} / (1 - R)$, R is the load ratio, σ_{rs} is the residual stress, K is the strength coefficient, and n is the strain hardening exponent. Crack initiation life and early growth, N_i , was found using the Smith-Watson-Topper's [24] formulation,

$$2N_i = [0.5 \Delta \sigma_{\max} / (\sigma'_f - \sigma_m)]^{1/b}, \quad (14.13)$$

where σ'_f and b are the fatigue strength coefficient and fatigue strength exponent, respectively. σ'_f and b are material properties which theoretically correspond to the region where crack initiation occurs. Residual stress, σ_{rs} , in the as-welded condition was assumed to be equal to the yield strength in the crack initiation region. The stress concentration factors reported by Radaj [40] and Kuang et al. [41] for tensile-shear specimens were adopted with an adjustment due to the different nominal stress definition.

Sheppard in 1993 [25] extended the application of the structural stress to include the fatigue crack propagation. The mode I SIF range was obtained in terms of the structural stresses due to the membrane load and bending moment. The structural stress relationship was later modified in 1996 to include the effect of forces normal to the sheet [20].

Some advantages and shortcomings are associated with this model. Both the crack initiation and propagation phenomena are explicitly accounted for, and the mode I SIF is updated during the crack propagation. On the other hand, the effects of modes II and III loading during crack propagation are ignored. Also, there is no

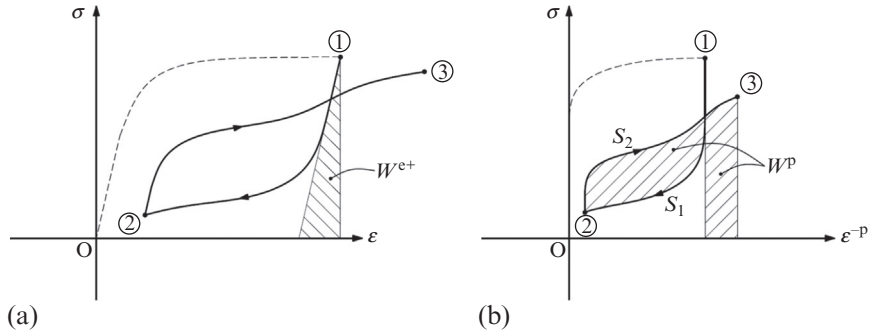
comprehensive formulation for calculating the stress concentration factor for spot-welds; this is a challenge in the application of Neuber's rule under a general load case. Furthermore, similar to other models based on the structural stress approach, effects of material anisotropy and yield asymmetry were not taken into account.

2.3 LOCAL NOTCH STRESS/STRAIN APPROACH

Some studies, in contrast to the fracture mechanics approach, consider a spot-weld as a blunt notch with a finite radius. Therefore, a detailed FE model with a fine mesh at the vicinity of the spot-weld is required. A measure of local notch strain at the spot-weld edge is often assumed to control fatigue failure in this approach. Local stress/strain values are calculated from an elastic-plastic FE simulation or from an elastic solution along with the Neuber's rule [23]. An advantage of this approach is that the effect of the spot-weld notch is considered. Also, cyclic characteristics of materials, for example, anisotropy and hardening asymmetry, may be accounted for. On the other hand, crack propagation is not considered in the local notch models. Moreover, complexities of the FE model preparation as well as the intensive FE calculations restrict the applicability of this approach for real-life problems. One of the most commonly referenced models in this group has been developed by Pan [27].

Pan in 2000 [29] showed that the Sheppard's model is capable of successfully predicting the fatigue life of steel spot-welds, as long as the sheets being welded have the same thickness; however, the Sheppard's model fails for dissimilar thicknesses. To tackle this problem, Pan in 2002 [27] proposed a model based on the local strain concept. A detailed FE model of spot-weld specimens with a nominal radius at the nugget edge was employed in this model. The nominal radius was selected according to experimental observations of the spot-weld cross-sections. Nonlinear geometry and nonlinear material properties were included in the FE analyses. Because the heat affected zone of spot-welds contains a gradient of material properties, this region was split into several subzones with corresponding nonlinear properties. Local stress and strain values at the spot-weld edge were obtained from three-dimensional elastic-plastic FE simulations. The critical point was identified as the location with the maximum local principal strain at the end of the first reversal. The maximum principal strain range at the critical point was used as the fatigue damage parameter. Maximum principal strain range, according to Pan [27], is the difference in the principal strain at the end of the loading and unloading reversals.

Behravesh in 2013 [10] proposed a local fatigue model based on strain energy. The advantage of energy-based fatigue models over stress- and strain-based models is that energy is a scalar quantity. Therefore, the energy values corresponding to different axial and shear components of stress and strain tensors can be simply manipulated without the concern of dissimilar nature or direction [10]. The total strain energy, as the damage parameter in this model, includes the plastic strain energy. The plastic energy dissipates in each cycle as a result of plastic deformation. To take the effect of mean stress into account, positive elastic strain energy was added to the plastic strain energy. Moreover, the positive elastic energy term helps to distinguish the data points in the high-cycle regime. Figure 14.5a schematically displays the equivalent stress versus total strain in the first three reversals.

**FIGURE 14.5**

Schematic illustration of (a) positive elastic energy density and (b) plastic energy density.

As depicted in this figure, the stress-strain response may constitute a nonclosing hysteresis loop. The positive elastic energy, W^{e+} , is defined as

$$W^{e+} = \frac{1}{2} \sigma_{\odot} \epsilon_{\odot}^e = \frac{\sigma_{\odot}^2}{2E}, \quad (14.14)$$

where $\sigma_{\odot}$ and $\epsilon_{\odot}^e$ are equivalent stress and elastic strain at the end of the first reversal, respectively.

The plastic strain energy, W^p , accumulated in the unloading (S_1) and reloading (S_2) reversals is calculated from:

$$W^p = \int_{S_1} \sigma d\bar{\epsilon}^p + \int_{S_2} \sigma d\bar{\epsilon}^p \quad (14.15)$$

The hatched area in Figure 14.5b represents the plastic strain energy. Note that Figure 14.5b is a plot of equivalent stress versus equivalent plastic strain.

Total strain energy, W^t , is the summation of the plastic strain energy and the positive elastic strain energy,

$$W^t = W^{e+} + W^p. \quad (14.16)$$

Total strain energy at the critical point is the fatigue damage parameter. The location with maximum equivalent plastic strain at the end of the first reversal is considered as the critical point.

3 CASE STUDY: FATIGUE LIFE ASSESSMENT OF AN AUTOMOTIVE DEMO-STRUCTURE

3.1 PROBLEM STATEMENT

This section discusses in detail the steps followed to assess the fatigue life of an automotive substructure (will be referred to as Demo-structure). The Demo-structure comprises three components: upper rail, lower rail, and shock tower, which are made

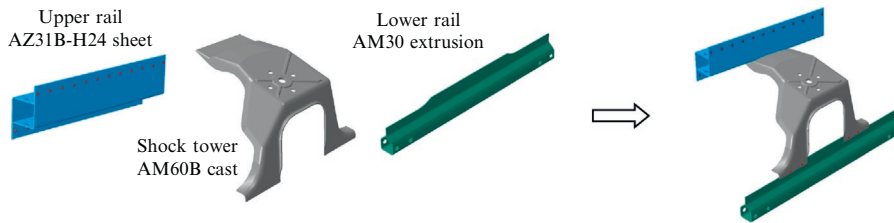
**FIGURE 14.6**

Illustration of the Demo-structure and its components [10].

of three different magnesium (Mg) alloys: AZ31B sheet, AM30 extrusion, and AM60B cast, respectively. The components of the Demo-structure are shown in Figure 14.6. This structure supports the loads transferred from the suspension system; therefore, fatigue is a serious concern in the structure and especially in its joints. For joining the components of the Demo-structure a modified form of the self-pierce riveting (SPR) was used which is introduced in the next section.

3.2 SELF-PIERCE RIVETING

The SPR is a technique mostly used for joining lightweight materials, such as aluminum and magnesium [42,43], and hence it is widely utilized by automobile manufacturers to join aluminum parts [44–46]. This joining technique is a cold forming process performed by punching a semi-tubular rivet such that it passes the top sheet and partially pierces the bottom sheet, as shown in Figure 14.7.

The SPR technique can be used for joining dissimilar materials and multilayer stacks, but it requires penetration into both materials, which means it is not suitable for brittle materials [42,48]. If the ductility of the materials being joined by the SPR is

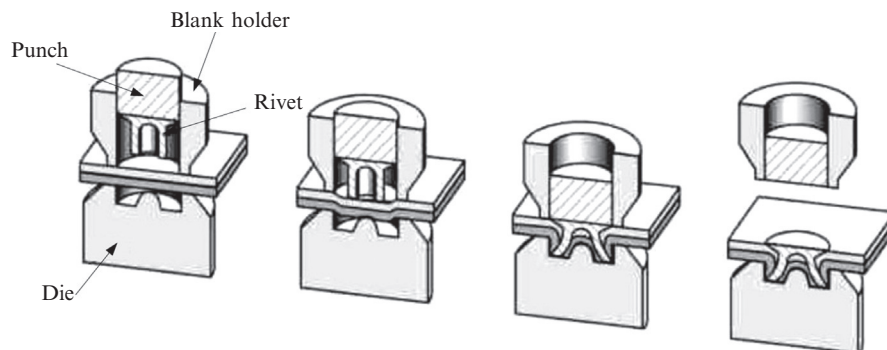
**FIGURE 14.7**

Illustration of the self-pierce riveting [47].

less than the required value, cracking might occur, especially in the bottom sheet, due to the severe plastic deformation.

Magnesium has hexagonal crystal structure which results in limited slip systems and poor ductility and formability at room temperature; therefore, SPR generally causes cracking in magnesium alloys [49]. The formability of magnesium improves at higher temperature because of the additional slip systems activated [50,51]. Durandet et al. [49] used a laser beam at the joint location and showed that it can provide enough heat to prevent initiation of cracks in AZ31 magnesium alloy. This method is called “laser-assisted self-pierce riveting” or LSPR, and it was employed for joining the components of the Demo-structure.

3.3 MATERIAL PROPERTIES AND MODELING

3.3.1 Material properties

As stated in Section 3.1, the components of the Demo-structure are made of three different magnesium alloys: AZ31B sheet, AM30 extrusion, and AM60B cast. In order to simulate the Demo-structure, the mechanical properties for these alloys were obtained from the literature [10,52,53]. The microstructure of the three alloys differ as a result of the different manufacturing process. The rolling and extrusion processes produce highly textured microstructures for AZ31B and AM30; therefore, material properties for these alloys are different under tension and compression. On the other hand, a randomly oriented crystal structure is observed for AM60B cast, which results in symmetric tension-compression behavior.

Figure 14.8 shows the monotonic behavior of the three alloys, in which RD, TD, and ED represent the rolling, transverse, and extrusion directions, respectively. Table 14.1 summarizes the monotonic properties of the three magnesium alloys in different directions.

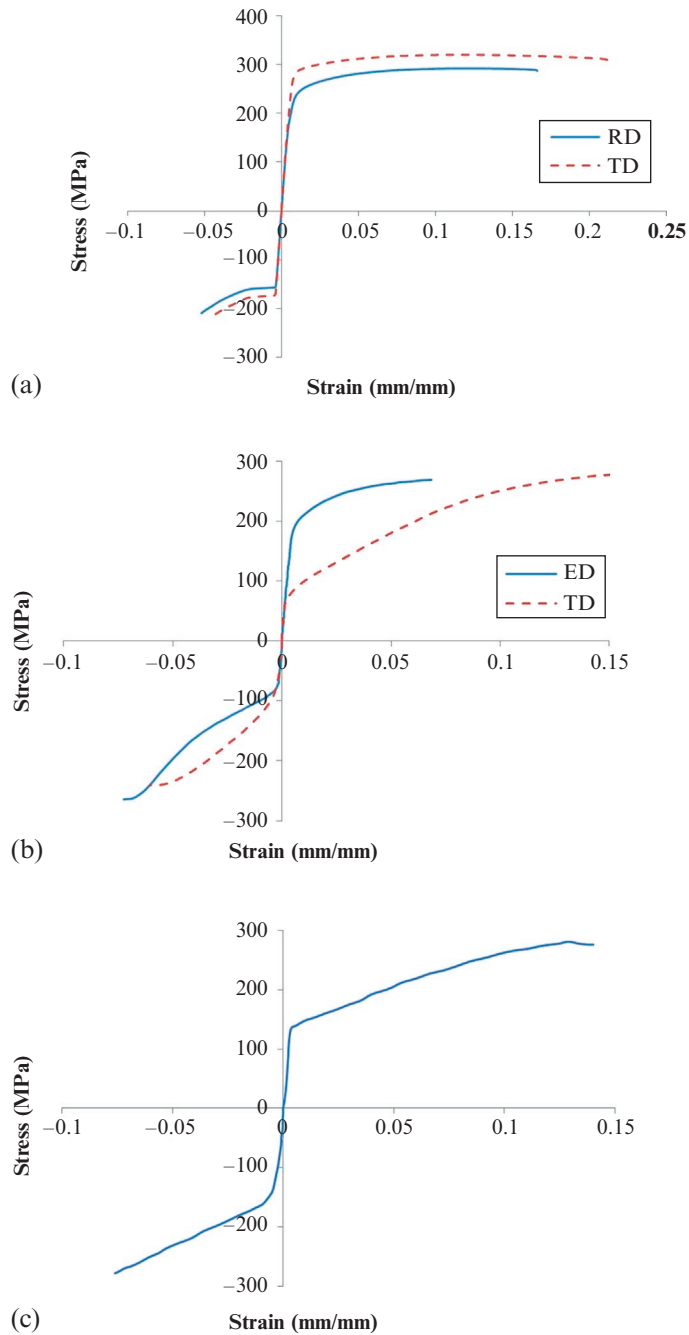
Similar to monotonic loading, cyclic loading experimental results for the three Mg alloys were obtained from the literature [10,54,55]. The stabilized hysteresis loops for the three alloys at different strain amplitudes are shown in Figure 14.9.

The strain-life curves were obtained for the three alloys under fully reversed loading [10,54,55] and are shown in Figure 14.10. In this figure, the fatigue properties correspond to RD for AZ31B, and ED for AM30 magnesium alloys.

The monotonic and cyclic properties of the rivets are unknown, and therefore the elastic steel properties, that is, $E = 210\text{ GPa}$ and $\nu = 0.3$ were used in the simulation.

3.3.2 Material modeling

After material properties are obtained from experiments or from the literature, the next step is to generate a FE model for the structure of interest and assign the cyclic stress-strain curves to the corresponding components. However, for the case of the Demo-structure the procedure is more complex than general problems. The wrought Mg alloys involved in the Demo-structure behave asymmetrically under tension and compression and this behavior could not be modeled by available material models in the Abaqus FE package [56]. Therefore, an asymmetric constitutive model was

**FIGURE 14.8**

Monotonic stress-strain curves for (a) AZ31B [10], (b) AM30 [52], and (c) AM60B [53].

Table 14.1 Monotonic Properties of AZ31B, AM30, and AM60B Magnesium Alloys

Alloy	Material Direction	Loading Direction	0.2% Yield Strength (MPa)	Ultimate Strength (MPa)	Fracture Strain (%)
AZ31B	RD	Tension	224	292	14
	RD	Compression	162	211	5
	TD	Tension	281	320	22
	TD	Compression	181	223	5
AM30	ED	Tension	196	270	7
	ED	Compression	84	265	7
	TD	Tension	76	279	15
	TD	Compression	119	241	6
AM60B	-	Tension	138	281	14
	-	Compression	146	278	8

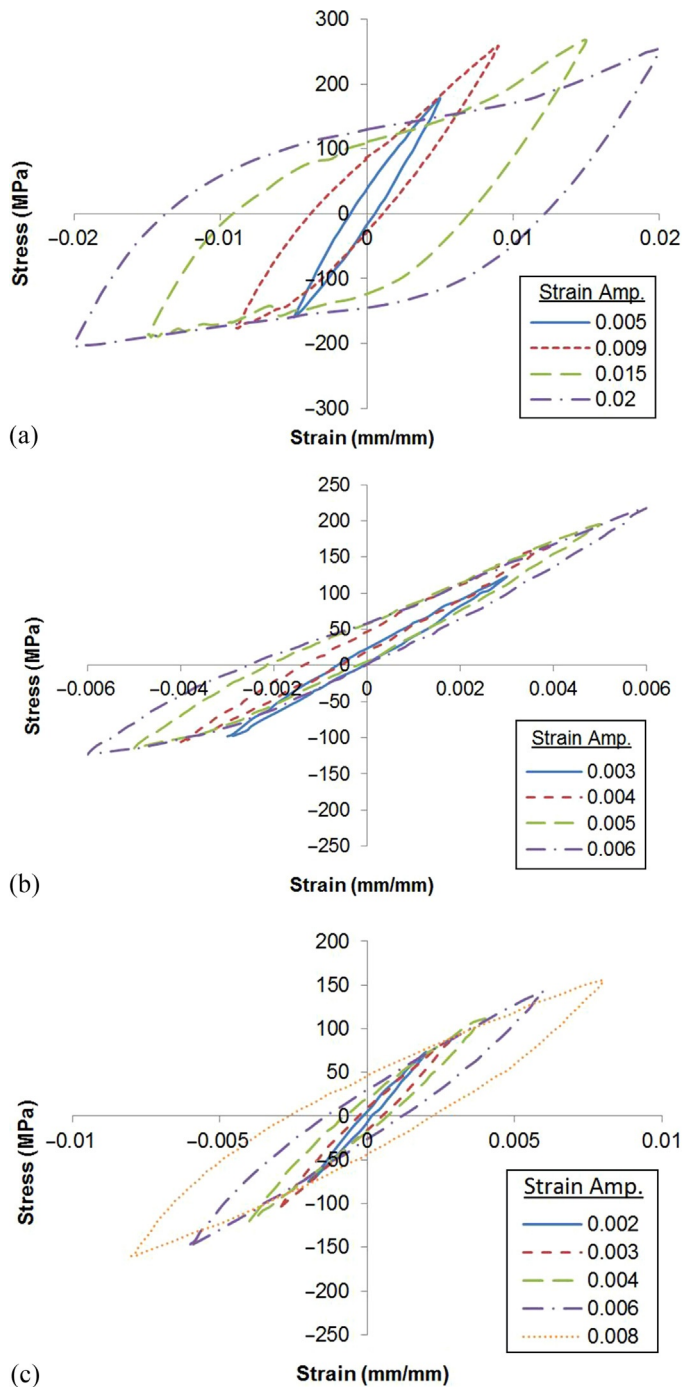
developed to predict the cyclic behavior of Mg alloys. The constitutive model was implemented into a user material (UMAT) subroutine to run with Abaqus/Standard. Because introducing the constitutive model and the numerical implementation is out of the scope of this book, these topics are not covered in this chapter and the reader is referred to the work by Behraves [10]. Figure 14.11 shows a typical comparison between the cyclic behavior obtained from the constitutive material model (UMAT) and the experimental results for AZ31B, AM30, and AM60B. This comparison demonstrates that the constitutive model is capable of predicting the cyclic axial behavior of Mg alloys.

3.4 FATIGUE MODELING

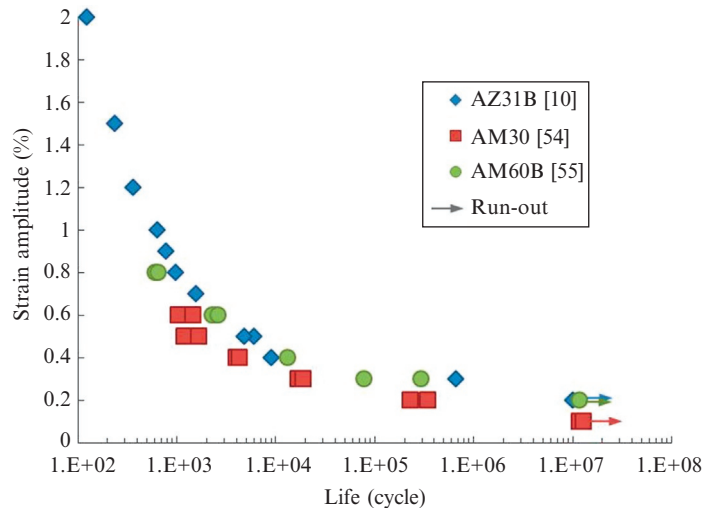
In the real-world problems, structures often include different components which are made of dissimilar materials and joined together with a consistent or various joining processes. To estimate the fatigue life of such a structure, a fatigue model must be chosen and the corresponding fatigue properties are needed for different materials and joints. The fatigue model proposed by Behraves in 2013 [10] was employed for this case study. As introduced in Section 2.3, this model takes total strain energy as the fatigue damage parameter. This fatigue model was applied to the fatigue test results of the three Mg alloys and the total strain energy was calculated for the stabilized stress-strain hysteresis loops at different strain amplitudes. Finally, the strain-life curves obtained from experiments were converted to energy-life curves, as shown in Figure 14.12.

Then, the fatigue life can be estimated from

$$N_f = P \cdot W_f^q \quad (14.17)$$

**FIGURE 14.9**

Stabilized cyclic hysteresis curves for (a) AZ31B [10], (b) AM30 [54], and (c) AM60B [55].

**FIGURE 14.10**

Fully reversed strain-life curve for different magnesium alloys.

where N_f is the fatigue life, and W^f is the total energy. P and q are the material constants obtained from the linear regression lines in Figure 14.12. These constants for the three magnesium alloys are presented in Table 14.2.

To predict the fatigue behavior of the Demo-structure, fatigue modeling of LSPR joints is also required. Because joints, in general, and spot joints, in particular, are locations of stress concentration, fatigue failure is very prone in these areas. Therefore, understanding and predicting the fatigue behavior of joints are critical for a reliable fatigue modeling of a welded structure. The next section deals with testing and fatigue simulation of LSPR specimens.

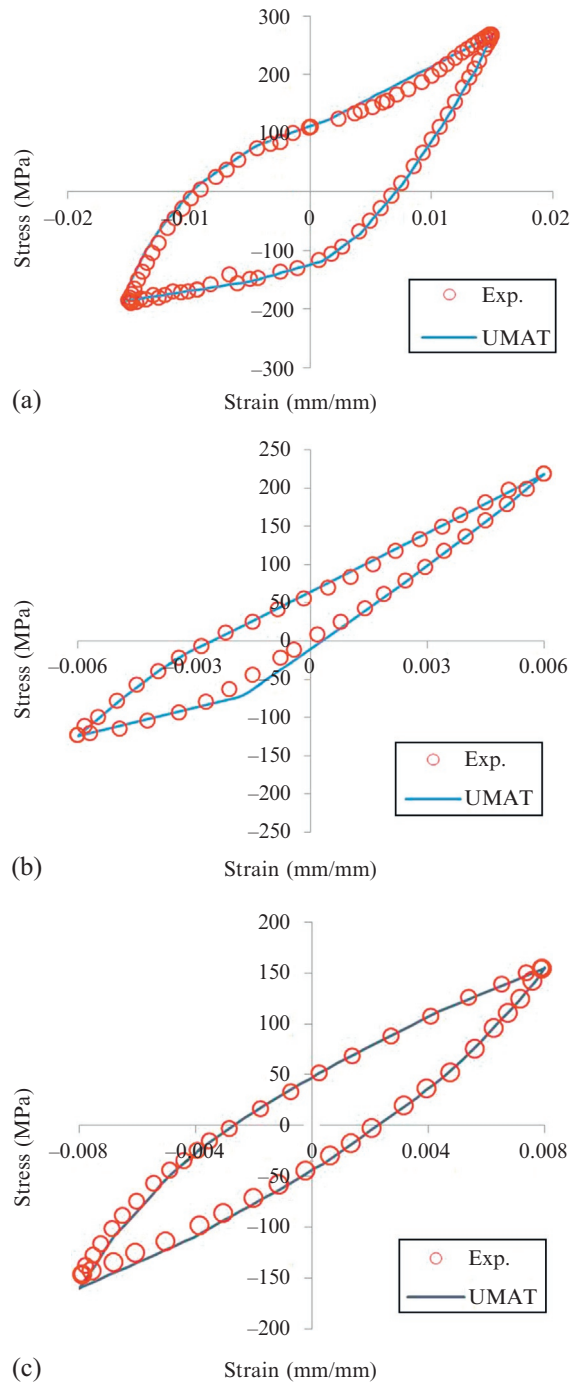
3.5 LSPR SPECIMEN

3.5.1 Specimen and testing

The fatigue strength of the LSPR joint was determined by testing specimens with the geometry shown in Figure 14.13.

Different sets of the LSPR specimens were produced with various material combinations as listed in Table 14.3. The preparation and testing of all LSPR specimens were conducted by various contractors for the US Automotive Material Partnership (USAMP).

Load-controlled cyclic tests with a load ratio of $R = 0.1$ were conducted to determine the fatigue strength. The results are shown in Figure 14.14.

**FIGURE 14.11**

Cyclic stress-strain response for (a) AZ31B under $\epsilon_{\text{amp}} = 1.5\%$, (b) AM30 under $\epsilon_{\text{amp}} = 0.6\%$, (c) AM60B under $\epsilon_{\text{amp}} = 0.8\%$.

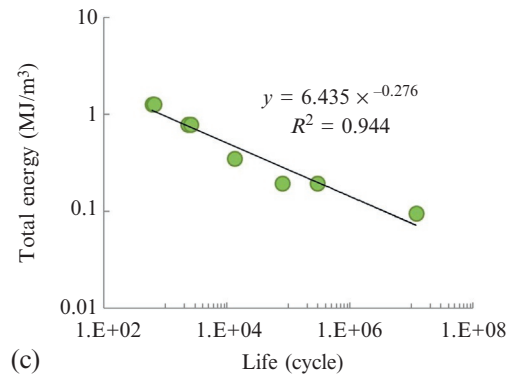
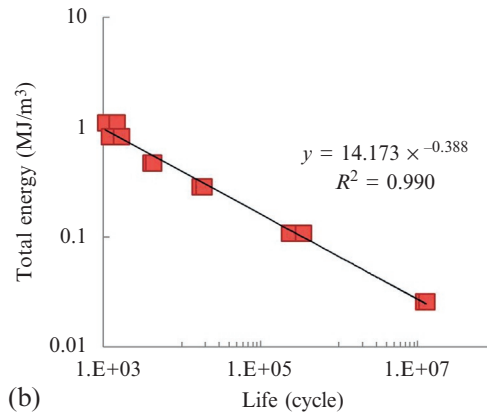
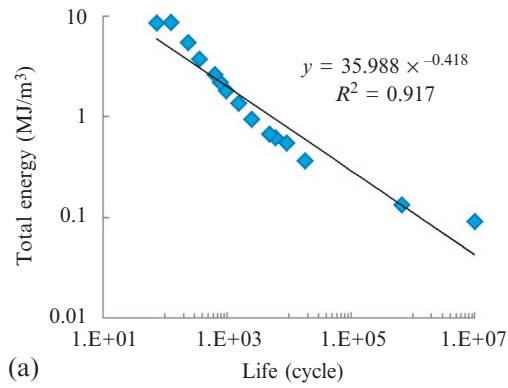


FIGURE 14.12

Total energy-life curves for (a) AZ31B, (b) AM30, and (c) AM60B [10].

Table 14.2 Constants of the Energy-Life Equation for Different Magnesium Alloys [10]

	AZ31B	AM30	AM60B
P	5257	925	844
q	-2.391	-2.576	-3.619

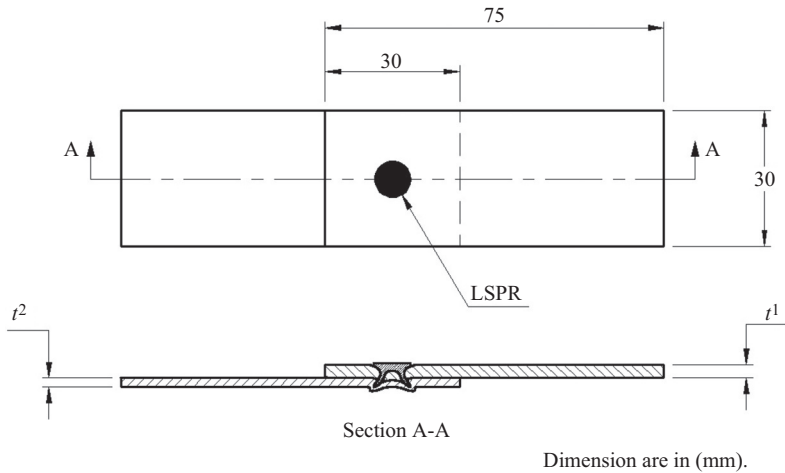
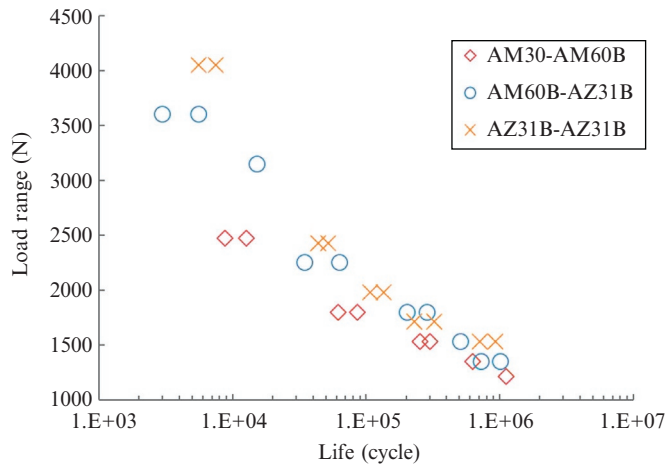


FIGURE 14.13

Geometry of LSPR specimen [57].

Table 14.3 Stack Composition of LSPR Specimens [57]

Specimen	Top Sheet—Thickness	Bottom Sheet—Thickness	Application
AM30-AM60B	AM30—2.5 mm	AM60B—2 mm	Lower rail to shock tower
AM60B-AZ31B	AM60B—2 mm	AZ31B—2 mm	Shock tower to upper rail
AZ31B-AZ31B	AZ31B—2 mm	AZ31B—2 mm	Upper rail (joining the two pieces)

**FIGURE 14.14**

Fatigue test results for different LSPR specimens [57].

3.5.2 FE modeling

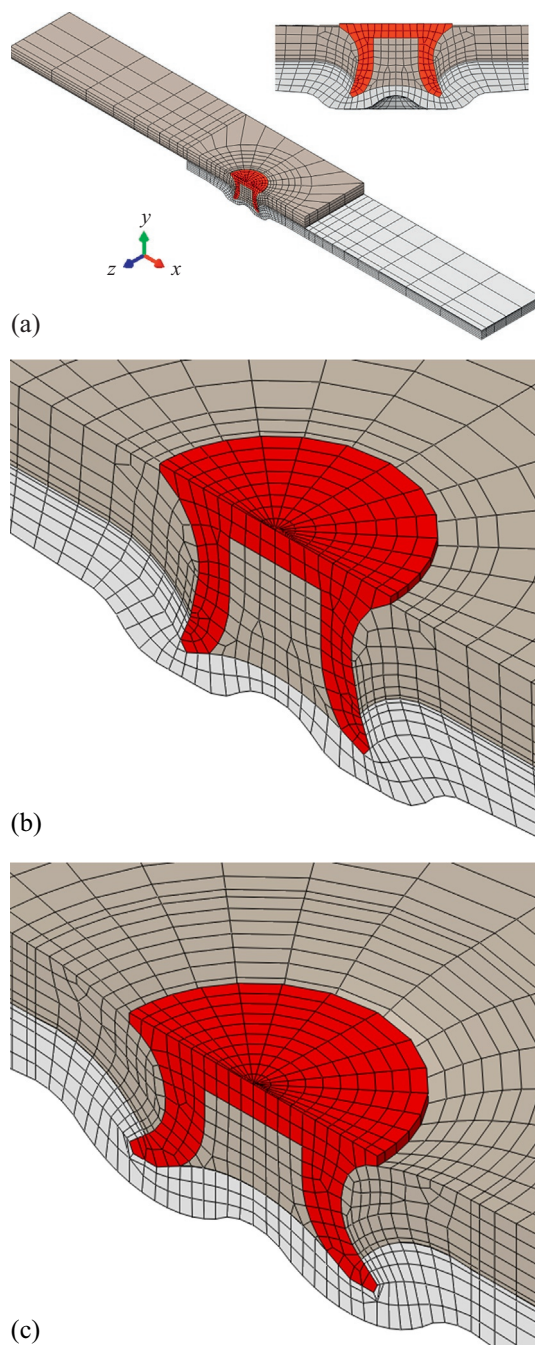
The LSPR specimens were modeled in the commercial FE software, Abaqus/CAE 6.10. Only half of each specimen was considered for the modeling considering a plan of symmetry, as shown in Figure 14.15. The elements employed in this model were eight-node linear brick elements with reduced integration and enhanced hourglass control (C3D8R elements). Fatigue properties of the base metals were assigned to the top and bottom sheets for each specimen (Table 14.3), and elastic properties for steel were used for the rivets.

The boundary conditions for the FE model including the end conditions and the symmetry conditions are shown in Figure 14.16.

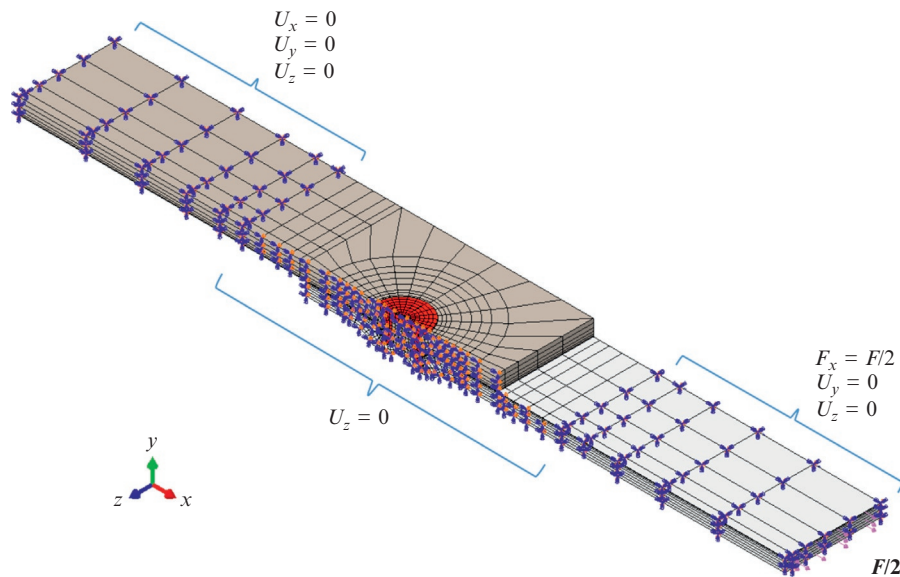
3.5.3 Simulation results and discussion

FE simulations were run in three steps representing three consecutive reversals: loading, unloading, and reloading. The total strain energy was the fatigue damage parameter and the critical point was identified as the location with maximum equivalent plastic strain at the end of the first reversal. The stress-strain solution was obtained from the FE simulation and the total energy was calculated at the critical point. This procedure was followed for all the data points presented in Figure 14.14. The total energy values were plotted against the observed fatigue life for the LSPR specimens, as shown in Figure 14.17.

The material constants for Equation 14.17, required for estimating the fatigue life of LSPR specimens, were obtained from Figure 14.17 and are presented in Table 14.4.

**FIGURE 14.15**

FE model for LSPR specimens (a) overall view of the half model. (b) 2.5 mm (top thickness) $\times$ 2 mm (bottom thickness). (c) 2 mm $\times$ 2 mm [10].

**FIGURE 14.16**

Boundary conditions for the half FE model of the LSPR specimens [10].

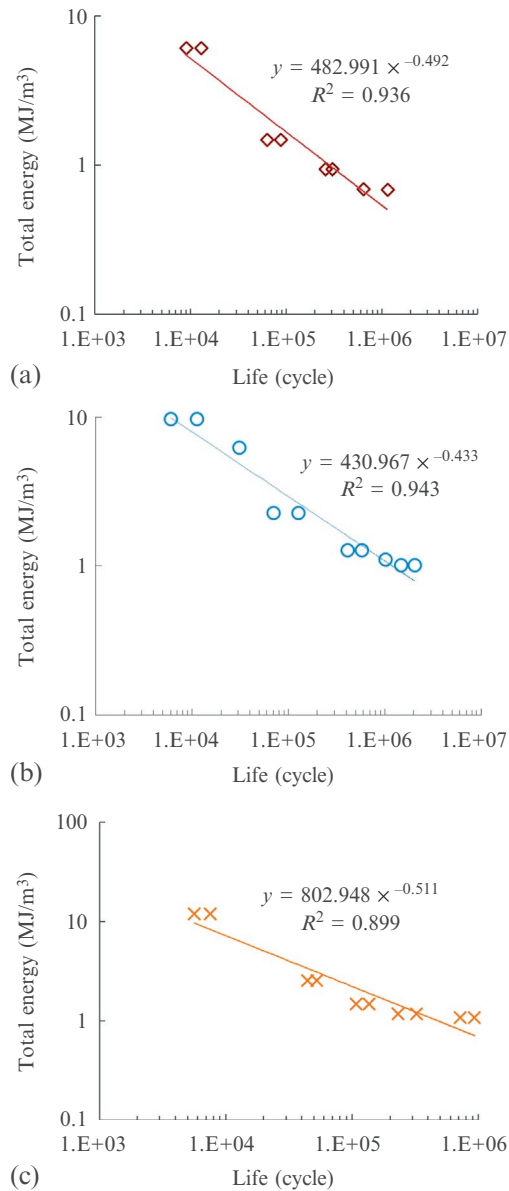
These constants are the material properties at the vicinity of rivets. One should note that the material behavior in this region is different than that in the base metal, due to the residual stresses induced by the LSPR joining process.

3.6 DEMO-STRUCTURE

The geometry of the Demo-structure was provided by the USAMP. The geometric model was further developed and the FE model was generated. In the FE model, the riveted areas were partitioned from the assembly to expedite the mesh generation phase, as shown in Figure 14.18.

One-eighth of each SPR joint was used for the FE modeling, which was then reflected to form the whole joint, as shown in Figure 14.19. As can be seen from this figure, mapped meshes were generated in the SPR joint regions with gradual refinement toward the interface to help capturing stress and strain distributions more accurately.

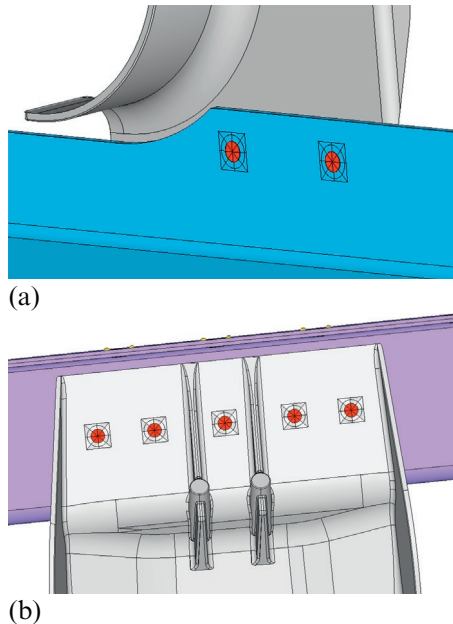
The FE model of the rivet was then copied to all similar locations and free meshing technique was used for the other regions of the Demo-structure due to the complexity of the geometry. Eight-node linear brick elements with reduced integration and enhanced hourglass control (C3D8R) along with 4-node linear tetrahedron elements (C3D4) were employed. Contact interaction was defined between matching surfaces to avoid intersection of different parts. The FE model of the Demo-structure is shown in Figure 14.20.

**FIGURE 14.17**

Energy-life curves for LSPR specimens. (a) AM30-AM60B, (b) AM60B-AZ31B, and (c) AZ31B-AZ31B [10].

Table 14.4 Constants of the Energy-Life Equation for Different LSPR Specimens [10]

	AM30-AM60B	AM60B-AZ31B	AZ31B-AZ31B
P	283,184	600,424	479,112
q	-2.031	-2.308	-1.956

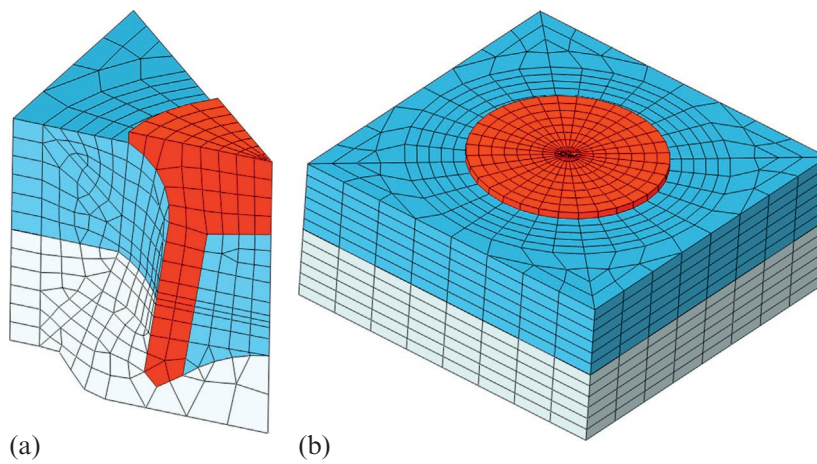
**FIGURE 14.18**

Partitioned SPR joints in the Demo-structure. (a) lower rail-shock tower joints, (b) shock tower-upper rail joints [10].

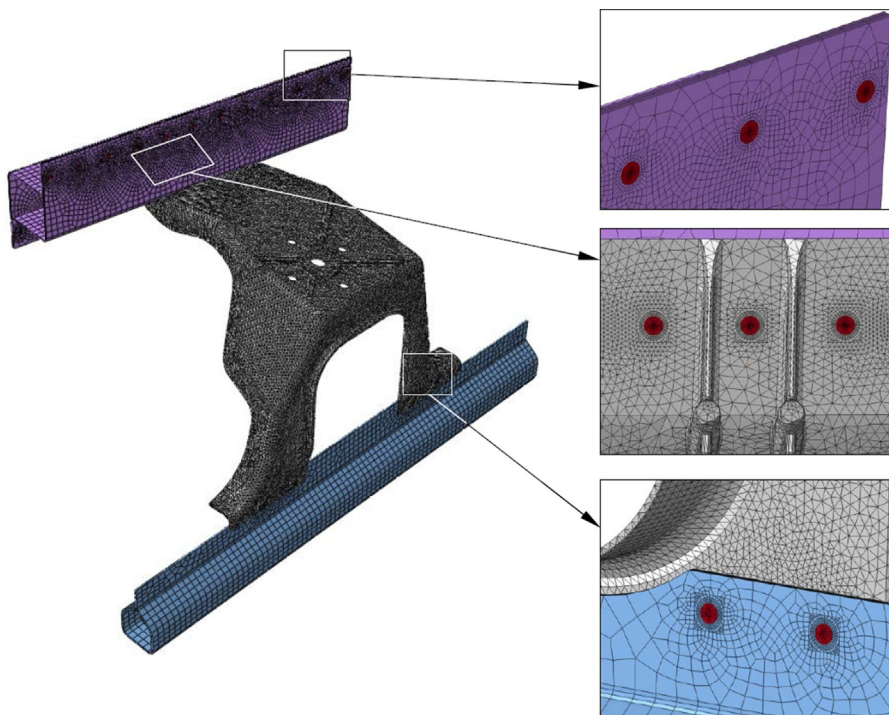
Several preliminary simulation on the Demo-structure model suggested that the SPR joints in the upper rail are not critical in all loading conditions of interest. Accordingly, these joints were ignored to save computation time, and the two sheets forming the upper rail were merged together.

The material constants presented in Tables 14.2 and 14.4 were assigned to the corresponding regions. Figure 14.21 shows the boundary conditions applied to the Demo-structure.

The loading condition was a maximum load of $F_{\max} = 4\text{ kN}$ and load ratio of $R = 0$ in the x-direction, as shown in Figure 14.21. The simulation was run in three steps to simulate three consecutive reversals, that is, loading, unloading, and reloading. To refer to different SPRs in the Demo-structure, the SPRs are numbered according to Figure 14.22.

**FIGURE 14.19**

FE model for a sample SPR joint (a) one-eighth model (b) whole model [10].

**FIGURE 14.20**

FE model of Demo-structure [10].

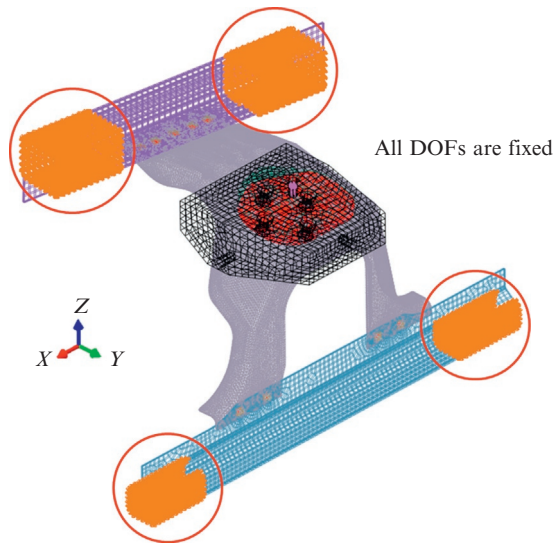


FIGURE 14.21

Boundary conditions of the Demo-structure [10].

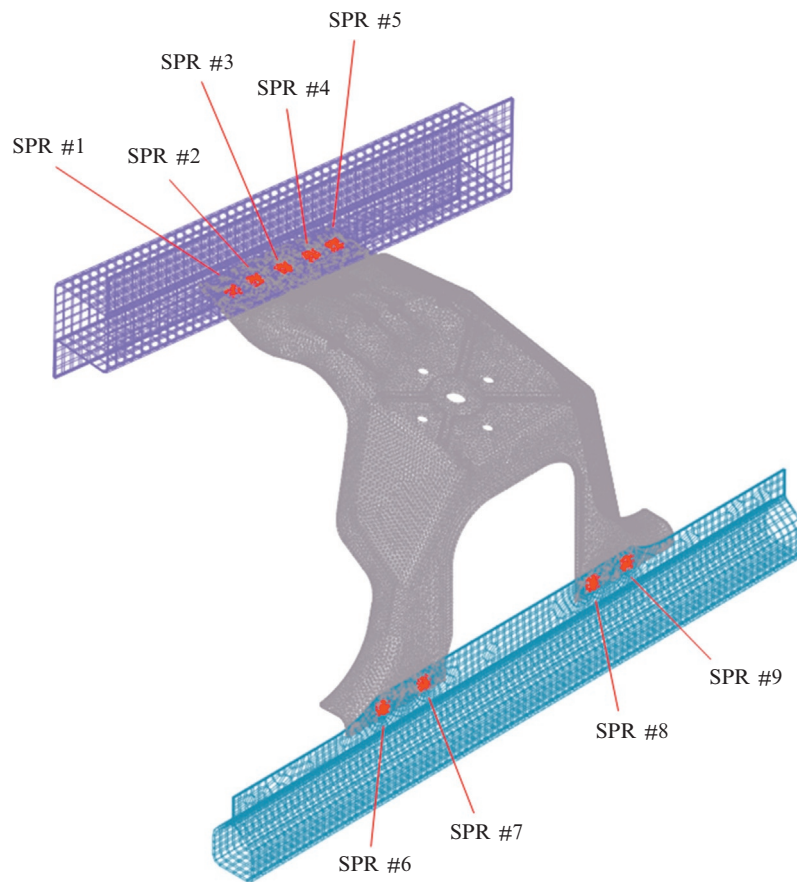


FIGURE 14.22

Numbering SPRs in the Demo-structure [10].

The fatigue simulation predicts that a fatigue crack initiates at the SPR #9 after 50,540 cycles. This prediction was in agreement with the experimental results. It should be noted that the fatigue model adopted in this case study predicts only the crack initiation life. Crack propagation phase must be dealt with separately, using a fracture mechanics based model.

4 CONCLUSION

In this chapter, fatigue models of welded joints were classified into different approaches and their features were explained. An automotive substructure was considered for the case study and its fatigue behavior was predicted using a fatigue model based on the local notch approach. Through the case study, the following conclusions were made:

- An unusual asymmetric shape of the hysteresis loop is the key feature of the cyclic behavior of wrought Mg alloys.
- The constitutive material model adopted in this study was capable of simulating the asymmetric cyclic hardening behavior of Mg alloys.
- Accounting for the asymmetric hardening behavior and using the local strain energy model resulted in a successful prediction of the failure location and crack initiation life.

ACKNOWLEDGMENTS

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Welding-associated failures in power boilers

15

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1 MATERIAL APPLICATIONS IN POWER BOILERS – INTRODUCTION

The function of a boiler is to produce superheated steam by transferring the heat produced by the combustion of oil or natural gas to water and steam flowing through the tubes that form the boiler. The super-heaters consist of banks of tubes suspended in the combustion gas flow path in the upper parts of the boiler. The steam flows from the super-heaters through the main steam header to the main turbine, where it drives the turbine-generator. The service conditions involved in boilers, which are mainly governed by alternating high temperatures and induced thermal stresses; residual stresses in welds; as well as the harsh corrosive environments, promotes failure causes such as thermal fatigue; stress-corrosion cracking (SCC); corrosion; and premature creep failures due to stress raisers at welding defects.

In high-temperature service (345-815 °C), and in the design of pressure vessels and piping, engineers and designers are confronted with the problem of selecting materials for a wide range of service conditions. The chromium molybdenum ferritic and austenitic stainless steels (SSs) are generally used for design temperatures above 425 °C [1]. In addition to service temperature, corrosion resistance, and fabricability, the following conditions should be considered in high-temperature applications: allowable maximum temperature and stress, type and value of load, expected life of the structure, and cost.

Austenitic SSs are generally used for design temperatures above 425 °C, the ASME code, section VIII, division 1, shows the SS grades that are suitable for the normal temperature range of usage (345-815 °C). Also some creep-resistant martensitic and ferritic SSs are used for high-temperature applications. The duplex stainless steels (DSSs) with both ferrite and austenite phases have gained increasing interest in recent years in power generation applications in chloride-containing solutions due to their better resistance to localized corrosion compared to single-phase austenitic SSs [2–4]. Welded DSSs are used at temperatures up to 250 °C. The increase in the thermal efficiency of fossil fuel-fired steam power plants, that can be achieved by increasing the steam temperature and pressure, has provided the incentive for the development of heat-resistant steels with excellent creep properties, as well as superior oxidation and corrosion resistance properties [5]. In the last two decades and to face up these requirements; several new Cr-Mo and 9-12% chromium steels were developed ranging from P11 (1Cr-0.5Mo) to P122 (12Cr-1Mo) [6].

Table 15.1 summarizes the main types of steels used in pressure vessels with their working conditions.

Table 15.1 The Main Types of Steels Used in Pressure Vessels with Their Working Conditions

Steel Type	Working Conditions
Ferritic stainless steels	430-510 °C
Austenitic stainless steels	595-815 °C
Duplex stainless steels	Temperatures up to 250 °C, with chloride-containing solutions
Creep-resistant martensitic steels	510-650 °C, with oxidation environments

Austenite is the predominant phase in austenitic SSs, while in DSS the chemical composition is tailored to produce a structure consisting of ferrite and austenite. Unfortunately, during welding of SSs many discontinuities may be produced. These discontinuities act as stress raisers that can lead to a decrease in the life of the weld. Moreover, many of the involved structures include straight runs of tubes, near bends, and welded tubes. High-strength creep-resistant steels are usually subjected to normalizing and tempering heat treatment prior to service and therefore, have microstructures of tempered martensite. Creep strength of these steels is improved by the martensitic lath structure, precipitation strengthening effects of M₂₃C₆ carbide and MX (M = Nb, V, Cr and X = C, N) carbo-nitrides, and the solid-solution strengthening effect of Mo and W in the matrix. The precipitation strengthening effect of MX carbo-nitrides is especially important, because its coarsening rate is small and the fine particle size is maintained for long-term service [7].

2 STAINLESS STEELS (SSs) AND THEIR WELDING CHARACTERISTICS

SSs are the most important family of corrosion-resisting alloys. These alloys acquire their SS properties from the presence of the chromium element. The corrosion resistance of these alloys allows them to find numerous applications in chemical plants, high-temperature applications, food processing equipment, and others.

The effect of chromium lays in the formation of a tight, impervious layer of chromium oxide, which protects the metal from the environment. The essential level of chromium in SS is more than 11%. At this level, the chromium oxide layer is continuous and provides full protection. The other alloying additions in SSs, such as nickel, provide additional desirable properties such as low-temperature toughness, improved ductility, and strength.

2.1 CLASSIFICATION OF STAINLESS STEELS (SSs)

The structure of SS is determined by the stability of phases, such as ferrite, austenite, and martensite in microstructure, and hence by alloy composition leading to these phase stabilities. In this respect, it is worth mentioning that Cr, Mo, Si, Ti, Nb, V, and Al are ferrite stabilizers, while Ni, Mn, C, N, Co, and Cu are austenite stabilizers. Based on the effect of these alloying additions, SSs are composed of four basic families of alloys: ferritic, martensitic, austenitic, and DSSs. In addition, there is a family of alloys relying on precipitation hardening to improve mechanical properties. The duplex alloys are designed to have nominally 50% ferrite and 50% austenite. They provide superior toughness, strength, and corrosion resistance than ferrite alloys and better resistance to SCC than the austenitic alloys, and thus they find increasing applications in oil, natural gas, and chemical industries. However, their use is limited to 250 (for welded structures) and to 300 °C for other structures in order to prevent precipitation of different undesirable phases.

Table 15.2 The Main Families of Steels with Their Main Microstructural Characteristics

Steel Class	Main Microstructural Characteristics
Ferritic	Ferrite phase
Austenitic	Austenite phase
Martensitic	Martensite
Duplex stainless steels	50% ferrite and 50% austenite
Precipitation hardening	Presence of MX carbo-nitrides

Table 15.2 summarizes the main families of steels with their main microstructural characteristics.

2.2 WELDING CONSIDERATIONS

Welding of SSs may cause a change in phase stability, precipitation of undesirable phases, or a delay in phase transformation due to welding heat cycle, in addition to welding residual stresses, which may cause failures in-service. The basic considerations of welding SSs as compared to carbon steels are discussed in the following sections.

2.2.1 Thermal properties

The melting temperature and the thermal conductivity are lower for austenitic SSs than carbon steels, while the coefficient of thermal expansion is higher in austenitic SSs than carbon steels. For example: for 304 stainless and carbon steels, these properties are 1425 °C, 16.2 W/(m. K), and $17.6 \times 10^{-6} \text{ }^{\circ}\text{C}^{-1}$; and 1510 °C, 46.0 W/(m. K), and $11.7 \times 10^{-6} \text{ }^{\circ}\text{C}^{-1}$, respectively. Thus, less heat is required to melt austenitic SS than carbon steel. However, due to the lower thermal conductivity, narrower heat-affected zone (HAZ), and slower cooling rates are encountered compared to carbon steels. This may cause distortion and buckling especially with the larger coefficients of thermal expansion of SSs [8,9]. More tack welding is required when welding austenitic SS to support the structure and decrease distortion. Heat input for welding DSSs should be as low as required for good welding quality and for avoiding harmful precipitates in the weld and HAZ.

2.2.2 Electrical resistivity

Electrical resistivities of SSs are generally higher than that of carbon steels, for example, four to five times larger than that of carbon steel at room temperature. This implies that higher heat generation is obtained in resistance welding and higher deposition rate in arc welding, that is, higher wire feed speed in submerged arc welding (SAW) and gas metal arc welding (GMAW).

2.2.3 Oxidation and depletion of chromium

Chromium is highly susceptible to oxidation and depletion due to the high chemical affinity of chromium to oxygen and carbon. This may lead to oxide inclusion defects and depletion of chromium from the weld metal and the HAZ, which deteriorates the corrosion resistance of welded joints. Therefore, there are some restrictions related to oxygen for welding SSs. Addition of stabilizing gas as oxygen or CO_2 in GMAW should be restricted to 2% of the shielding gas to avoid loss of alloying elements. The fluxes of SAW and shielded metal arc welding (SMAW) should have low oxygen levels to reduce oxidation and fluoride additions to facilitate chromium oxide fluxing. During the root pass welding of pipes, it is essential to protect the unshielded side by inert gas purging, this helps to avoid the loss of corrosion resistance at the inner side of the weld [8,10]. High heat inputs should be avoided to prevent chromium depletion in HAZ during slow cooling after welding of austenitic SSs.

After this concise review on some basic considerations of welding austenitic and DSSs, it might be useful to discuss some of the welding parameters and the potential welding problems and their remedies.

2.3 REVIEW OF WELDING PARAMETERS RELEVANT TO STAINLESS STEELS

The combination effect of heat input and heat treatment on the microstructure and mechanical properties of weld metal and HAZ in SSs and Cr steels has a major role on material performance. This combined effect shows on the resulted microstructure and formed phases and carbides as well as their stability. Accordingly, the final creep-resistant prosperities and SCC resistance are affected. Stringent control of welding conditions as per process qualification records requirements is an essential process for controlling the weld quality.

However, daily failures during service are reported and investigated from the chemical and energy plants [11,12], many of which prove to be originating from welded joints as they act as critical places where high residual stresses combined with heterogeneous structures exist so they act as potential sites for failures by stress-driven or environmental-driven mechanisms[13].

Though welding of austenitic SS types has reached a mature state of art, it has been shown that two ferrite morphologies precipitate at the austenite matrix of AISI/SAE 304 SS welded with an electrode type 308L using SMAW, GMAW, and flux-cored arc welding (FCAW) processes (skeletal and lathy type, the latter being characteristic of higher ferrite content). It has been also shown that (unlike FCAW) SMAW and GMAW cause a variation in grain size, as well as the appearance of exogenous phases [14].

DSS have a two-phase microstructure (austenite and ferrite) resulting their higher strength, higher resistance to intergranular corrosion, and lower cost compared to austenitic SSs. However, their difficult thermomechanical processing conditions limit their applications. Also, they are sensitive to heating and cooling conditions during welding and deleterious phases such as (σ phase and Cr_2N) may appear during welding of these steels. The Cr_2N phase is reported [2] to be the main precipitate found in the HAZ of

welded joints. The conditions that favor the formation of austenite, such as high nitrogen content and low cooling rate, minimize the formation of Cr_2N precipitates, and increase the volume fraction of austenite. During cooling after welding of DSS the δ -ferrite transforms to austenite; the greater the cooling rate after welding the lower will be the austenite content, which deteriorates the tensile and corrosion properties of weld metal and HAZ. In order to increase austenite content in the weld metal, Ni-enriched filler metals (9% Ni) are used. The structure in the HAZ would accordingly exhibit increased δ -ferrite/austenite ratio compared to the base metal due to either fast cooling rates from the solution annealing temperature 1050 °C or to decomposition of the austenite to ferrite during holding at the temperature $\cong 600$ °C [15]. Moreover, DSS are prone to age hardening and embrittlement at virtually all temperatures above 300 °C. At higher temperatures above 475 °C rapid aging results from the precipitation of various intermetallic phases (σ , χ , R , and G phases) [16].

2.4 WELDING PROBLEMS

The common welding problems encountered in SSs are sensitization, SCC, hot cracking, HAZ cracking, weld metal solidification cracking, and fissuring. These problems are discussed briefly in the following sections.

2.4.1 Sensitization

This problem can cause a rapid attack at the welding position. When the steel is heated at the temperature range of 500–850 °C, the carbon combines with chromium and forms carbides. These carbide particles form preferentially at grain boundaries, causing the depletion of chromium at the surrounding area. This results in a local loss of corrosion resistance. This process is called sensitization. This kind of damage takes place during welding at the HAZ, where the peak temperature reaches 600–850 °C, and therefore a localized damage called weld decay occurs in the HAZ, see Figure 15.1. The severity of sensitization depends on the carbon content of the base metal, the temperature, and the time of heating.

To avoid sensitization, three approaches are useful: (i) use of low carbon alloys such as 304L, 316L, 317L, these alloys do not have sufficient carbon (<0.03%) to

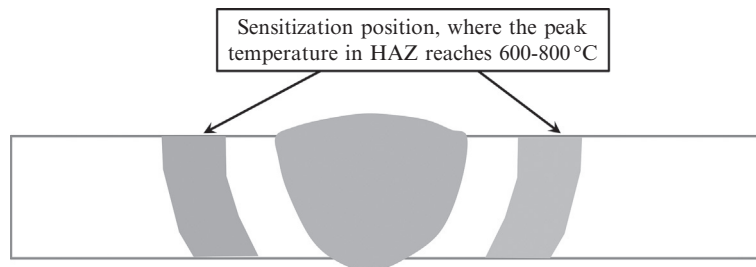


FIGURE 15.1

Sensitization in the heat-affected zone [8].

react with chromium; (ii) use of stabilized SS grades (e.g., 321 and 347), which contain alloying additions with higher affinity to carbon than chromium, such as titanium and niobium; and (iii) carrying out a solution treatment for the welded joint at a temperature range of 1050–1100 °C, this treatment helps to dissolve any chromium carbide in the joint and restores the corrosion resistance [8,17].

2.4.2 Stress-corrosion cracking

This damage takes place as a result of tensile stresses combined with corrosive conditions [18]. Environments with halogen ions such as chlorides are commonly attributed to these failures. The welding residual stresses play major role in these failures. As a consequence, stress relieving may constitute a remedy for this problem.

2.4.3 Cracking

There are several causes of cracking in SSs welds. Hydrogen cracking is common in ferritic and martensitic grades. Stabilized austenitic SSs are prone to HAZ cracking after welding and reheat cracking in-service. The first is due to the formation of low melting point compounds at grain boundaries, while the second mechanism is due to precipitation hardening inside the grains. Both mechanisms are promoted by the stabilizing element [8].

2.4.4 Solidification cracking

Typical solidification cracking appears at the centerline of weld metal. This problem is caused by two factors: the presence of sulfur and phosphorus impurities, and the rejection of these elements to grain boundaries, where it forms a low melting phase, which may tear out due to tensile stresses of shrinkage late during solidification. The mode of solidification also affects this phenomenon, so that crack-free weld can be obtained when the metal solidifies as ferrite-austenite rather than austenite-ferrite, since ferrite has high solubility of sulfur and thus, little rejection to grain boundaries occurs[19].

2.5 POST-WELD HEAT TREATMENT

With regard to boiler steels, the final heat treatment for T/P91, T/P92, and T/P911 consists of normalizing and tempering, an austenitising temperature of about 1060 °C is adopted for hardening. A fully martensitic structure is obtained upon cooling to room temperature over a wide range of cooling rates. Welding of these steels is similar to welding hardenable alloy steels, since complete transformation to martensite during air cooling after welding occurs, therefore, the specification of preheat and post-weld heat treating is a major requirement for these steels. A detailed investigation on this steel is given in Ref. [20]. P91 can be welded satisfactorily by many processes including manual metal arc, submerged arc, and gas tungsten arc welding. Postweld heat treatment (PWHT) is necessary for tempering the martensite formed during welding, and many investigations have highlighted the need for optimizing the PWHT temperature and time as well as filler material composition [21,22].

On the other hand, due to the heat input during welding, not only different microstructures are obtained within the weld but also microstructural changes take place in a small area besides the fusion line. It has been shown that there is a great influence of heat input and heat treatment on the mechanical properties of P91 weld joints, where the best combination of properties was obtained in low to medium range of heat input, between 1 and 2.8 kJ mm⁻¹ for both treatments, and a 90% increase in time to rupture was observed for normalizing/tempering treatment compared to subcritical post-weld treatment [23,24].

3 STRESS-CORROSION INDUCED FAILURES

3.1 PITTING CORROSION OF REBOILER 304L TUBES RESULTING FROM WRONG MATERIAL SELECTION

Some case studies, with special relevance to failure by stress-corrosion induced failures will be discussed hereafter. Since recently a detailed study on failures associated with ferritic steels has been presented [25], this work will focus on failure cases on austenitic and DSS grades (304, 316 and duplex steel) that failed due to combined environmental and stressing conditions by corrosion- and/or mechanical-induced mechanisms.

A number of the reported failures on parts made from SSs have been attributed to either pure corrosion or combined stress-corrosion causes [11,26–28]. However, failures may be attributed to pure mechanical causes only; the failure of two centrifugal pump shafts (one made from 316 austenitic and the other from DSS) during use in a Brazilian petroleum refinery were attributed to fatigue [13], where in both cases, the failure started with crack nucleation and was promoted by the presence of nonmetallic inclusion particles, which appeared in the austenitic SS as a result of the early stages of steelmaking process (in the 316 SS shaft), and could not be attributed to the material properties in the case of DSS (where the failure started at an edge, where a nut locked the rotor in the shaft).

Pitting corrosion and/or SCC were found to be the cause for a number of the reported failures in SS parts in both welded or none-welded parts, in various structures made of 304 SSs working in aggressive environmental conditions [29–32].

Pitting corrosion, or pitting, is a form of extremely localized corrosion that leads to the creation of small holes in the metal. The driving power for pitting corrosion is the lack of oxygen around a small area. This area becomes anodic while the area with excess of oxygen becomes cathodic, leading to very localized galvanic corrosion. The corrosion penetrates the mass of the metal, with limited diffusion of ions, further pronouncing the localized lack of oxygen. The mechanism of pitting corrosion is probably the same as crevice corrosion [33]. This kind of corrosion is considered more dangerous than uniform corrosion as it causes little loss of material with small effect on its surface, while it damages the deep structures of the metal. The pits on the surface are often covered by corrosion products and the pits penetrate through the

thickness at faster rates than general corrosion. Pitting corrosion is usually initiated by a number of factors depending on the working environment. The pits usually initiate at surface defects (scratches, local changes in composition, or damage to protective coating). Pitting corrosion may also be caused by bacteria in promoting environments such as the oil industry. Microbiologically influenced corrosion caused by sulfate-reducing bacteria has been found to be a serious threat to pipelines. The presence of oxygen (even at low levels 0.1-0.2 ppm in pipeline water) promotes bacterial corrosion. The presence of CO₂ also causes severe corrosion, which is sometimes called “mesa” corrosion [34].

A failure study on two reboiler 304L tubes from two trains at Obaiyed Field was conducted at the Faculty of Engineering Cairo University [35]. The steam reboilers consist of a large number of tubes bent in U shape supported by tube sheets, all being contained in a large shell. The tubes carry hot oil at about 260 °C under pressure of 12 bar and the outer diameter of the tubes (shell side) is in contact with water. The temperature and the pressure of the water were 150 °C and 5 bar, respectively. The water analysis obtained from BAPETCO reports showed that the chloride content was in the range of 10-70 ppm for train E2613 and from 8 to 62 ppm in train E2623. This range of chloride content is relatively high for austenitic SS 304, especially at high working temperature of this equipment. The data indicated that the chloride content in the steam generators feed water was around 15 ± 5 ppm and around 60 ± 5 ppm at the outlet. The data also showed that the pH ranged from 5.8 to 10.4.

Table 15.3 shows the chemical analysis of the tubes’ steel from which it is confirmed to be accepted as 304L. This steel is known to have the lowest pitting resistance among austenitic SSs. It may be used in environments containing very low chloride contents especially at relatively high temperatures. High chloride content can cause pitting. High chloride content and high tensile stresses can cause SCC in this type of steel. The tensile strength as shown in Table 15.4 also confirms that the steel complies with standard requirements.

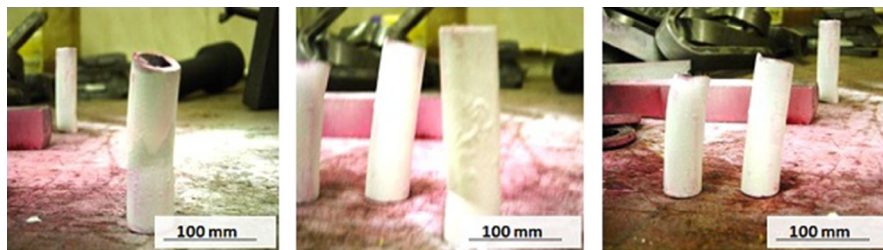
The two tubes were investigated visually; where one tube (E2613) showed pitting to a greater degree than the other one (E2623). The surfaces of both tubes were

Table 15.3 Chemical Analysis of Tubes’ Steel

Sample	C	Si	S	P	Cu	Mn	Cr	Ni	Mo	V	Al
E2613	0.03	0.49	0.009	0.036	0.11	1.69	20.1	10.37	0.27	0.07	0.004
E2623	0.027	0.47	0.007	0.034	0.11	1.75	19.8	10.22	0.21	0.06	0.006

Table 15.4 Tensile Strength of Tubes’ Steel

Specimen Code	0.2% Yield Strength (N/mm ²)	Ultimate Tensile Strength (N/mm ²)	%Total Elongation
E2623	271	574	47

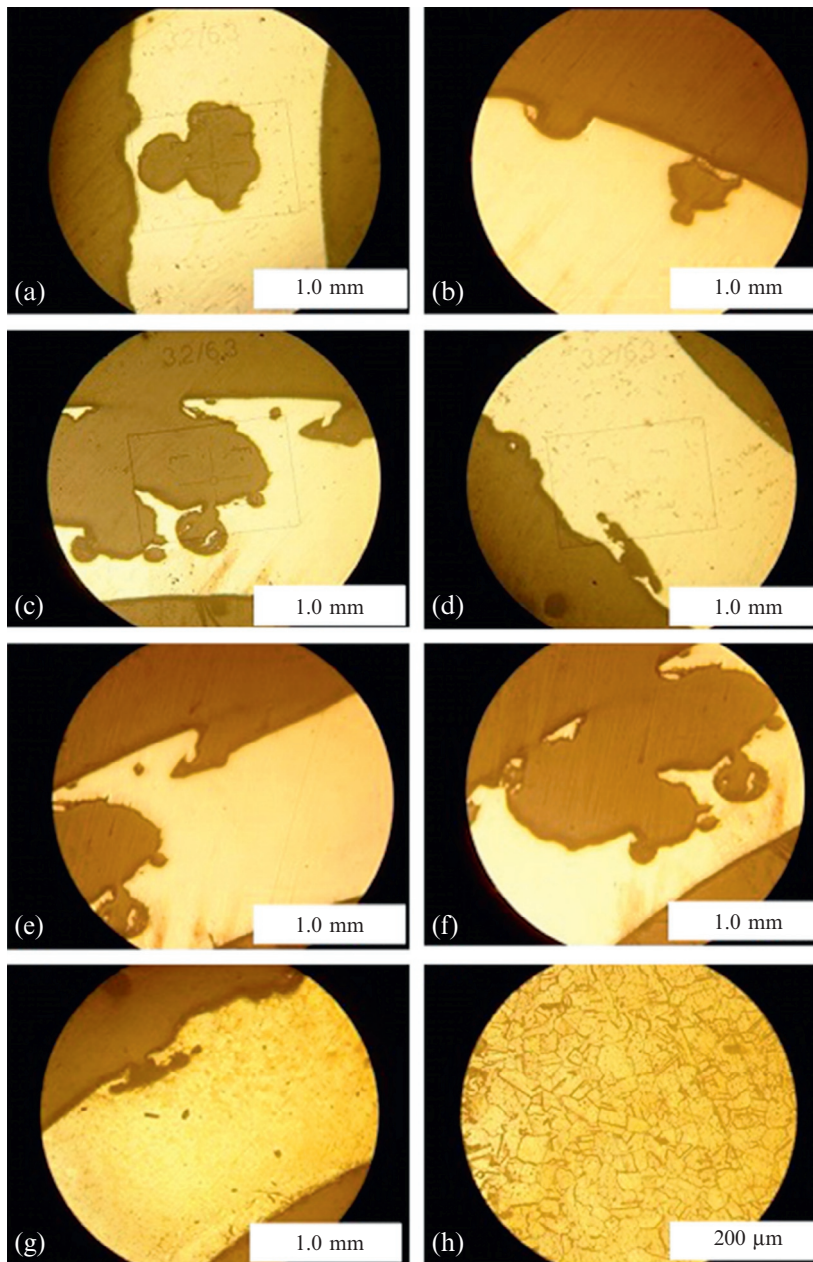
**FIGURE 15.2**

Liquid penetrant test of the failed tubes [35].

covered with black scale. No cracks were observed visually. For visual and penetrant tests, specimens were taken from the straight areas of the tubes as well as from the bent areas and the areas of the contact between the tubes and support plates. The penetrant test was carried out using solvent removable visible penetrant. No cracks were observed as seen from Figure 15.2.

Visual and macroscopic study of the unetched surfaces of the failed tubes (Figure 15.3) revealed the presence of various forms of subsurface and horizontal pits, where some were very deep and penetrated more than 75% of the thickness in one of the tubes, while others were narrow at the surface and wide under the surface. The pits were filled with a black dense residue. No cracks were found in 25 investigated cross-sections taken from the tubes even below the pits. Etched specimens showed normal austenitic microstructure under and around the pits (Figure 15.3). The chemical analysis of the scale is shown in Table 15.5, from which it is seen that the scale contains mainly oxides of iron, vanadium, and chromium. Vanadium and chromium oxides probably come from the process conditions.

The absence of cracks (visually or by penetrant test) indicated that SCC was not the reason of failure. This may be attributed to the very low stresses acting on the tubes at the working pressure of 12 bars (about 6.0 N/mm^2), which is below the threshold stress for SCC. The presence of the dense black layer on some of the pits suggested possibility of the occurrence of crevice corrosion initiating at the contact areas between the tubes and the support plates. Specimens from contact areas between the tubes and support plates were examined using liquid penetrant test, but no evidence for such pits was revealed at the tube-plate contact areas. Accordingly, the failure reason was suggested to be pitting corrosion. The high chloride content combined with stagnant conditions induced pitting of the SSs. Once the pits were initiated, the liquid trapped inside the pit became denser and more acidic than the bulk, and corrosion was accelerated, extending partly toward the inner surface of the tube and partly in a downward direction. However, in some locations evidence for crevice corrosion was found beneath the dense oxide scale that covers the outer surface, this suggests the importance of periodic cleaning of the tubes. A previous study on corrosion of steel weldments in water desalination plants has revealed that left-over slag on weld surfaces can produce an initiation point for crevice corrosion [26].

**FIGURE 15.3**

Unetched specimens from E2613, (a–f), and etched specimens from E-2613 at two different magnifications (g, h), 50 $\times$ [35].

Table 15.5 Chemical Analysis of Elements of the Scale on the Tube (E2623)

C%	V%	Mn%	Cr%	Ni%	Fe%	Mg%
1.4	28.2	0.54	6.7	2.4	20.64	0.06

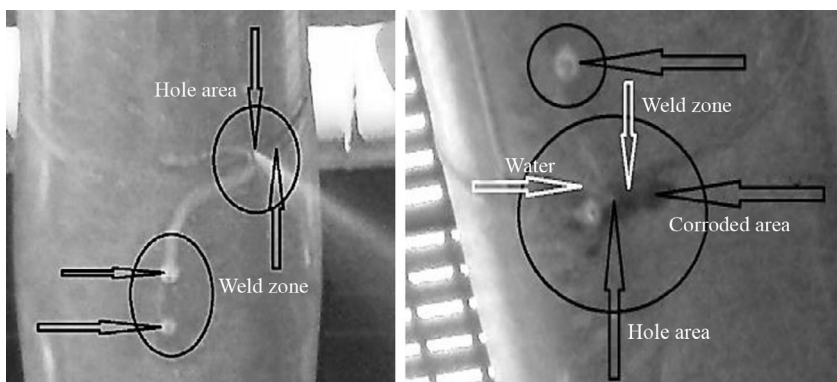
Pitting corrosion in steels is usually promoted by surface deposits that set up local concentration cells and dissolved halides that produce local anodes by rupture of the protective oxide film. Anodic corrosion inhibitors, such as chromates, can cause rapid pitting if present in concentrations below a minimum value that depends on the metal environment combination, temperature, and other factors. Pitting also occurs at mechanical ruptures in protective organic coatings if the external environment is aggressive or if a galvanic cell is active. Pitting normally occurs in stagnant environments. With corrosion-resistant alloys, such as SSs, the most common cause of pitting corrosion is highly localized destruction of passivity by contact with moisture that contains halide ions, particularly chlorides. Chloride-induced pitting of SSs usually results in undercutting, producing enlarged subsurface cavities, or caverns [36].

Since decreasing the chloride content below the current levels is impractical for such applications, it has been recommended to consider replacing the austenitic SS grade 304L by a SS of a higher pitting resistance number (e.g., 316L or 904L or even duplex steel grades).

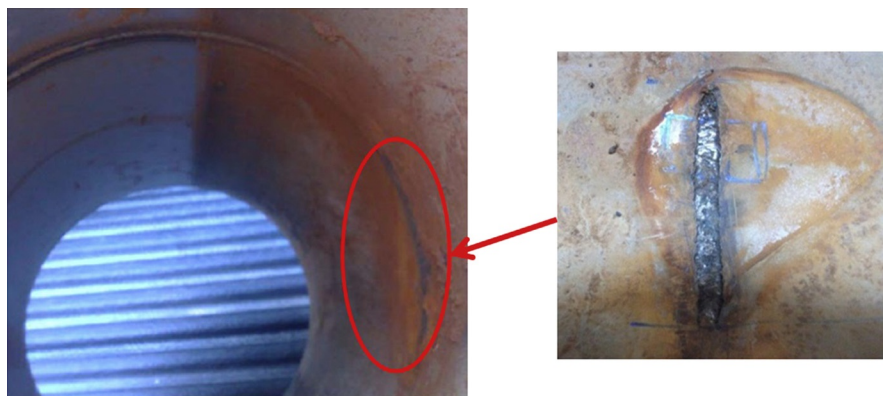
3.2 FAILURE OF A SUPER-AUSTENITIC STAINLESS STEEL FIREFIGHTING WELDED PIPE LINE RESULTING FROM WRONG WELDING PARAMETERS

A failure investigation on super-austenitic stainless steel (SASS: UNS S31254) firefighting pipe line was reported recently [37]. SASSs are a special class of austenitic SSs with higher Mo and Cr content for high chloride-containing environments. The firefighting pipe was welded by GTAW process using a nickel base filler metal (ER NiCrMo3). The firefighting network was established in the oil and gas field to provide the fire protection through piping seawater as firefighting water. Leakage of the firefighting was observed at the weld line and/or adjacent to the weld zone (i.e., at HAZ) after the firefighting network was pressurized and put in-service as shown in Figures 15.4 and 15.5. The failure mechanism was identified to be pitting corrosion along the weld joint. Microstructural examination of the failed area showed that the failure started at the fusion zone between the weld metal and the HAZ as shown in Figure 15.6. This area is called “unmixed zone” (UMZ) and has a microstructure of dendrites of austenite similar to that of an autogenous weld [38].

EDX analysis showed that Mo and Cr levels varied across the joint. Mo has a partitioning coefficient less than 1; hence, it tends to segregate during solidification to the grain boundaries. Cr, also, segregates during solidification in a manner similar to that of Mo but with lesser degree. Thus, the core of the dendrites within the UMZ

**FIGURE 15.4**

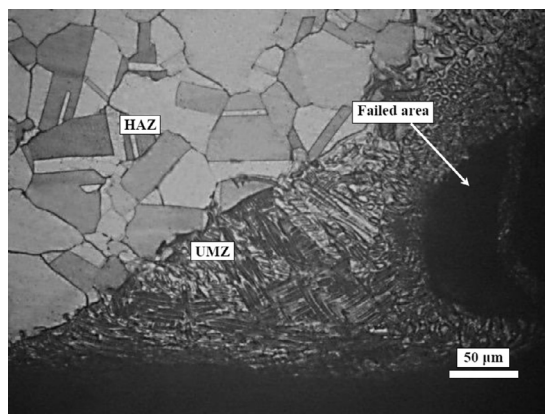
Water leakage in the firefighting pipe due to corrosion holes at the weld line and adjacent to the weld line [37].

**FIGURE 15.5**

The internal surface of the failed firefighting pipe showing weld corrosion [37].

becomes depleted in Mo. The molybdenum content in the dendrite cores can be as low as 4.2 wt.% leading to occurrence of localized attack in harsh environment [39]. This problem arises when SASS are welded using Ni-base filler metal such as ER NiCrMo3. The root cause of the failure case was the use of improper welding condition.

Accordingly, further investigation was conducted on the proper welding conditions for welding SASS with high Cr and Mo contents. GTAW with two different fillers were compared; namely: bead-on-plate using Ni-base filler metal (ER NiCrMo3) (similar welds) and fusion welding using Incoloy 28 filler metal (dissimilar welds).

**FIGURE 15.6**

The microstructure at the failed area. The failure starts at the fusion zone (UMZ) between the weld metal and the HAZ [37].

For both cases, welding heat input was varied within the range from 1 to 5 kJ/mm. The effect of welding heat input on the microstructure and microsegregation in the UMZ, and the corrosion properties was investigated for both cases. The increase in heat input to 5 kJ/mm led to slight decrease in the Mo microsegregation in the UMZ and appreciable amount of NbC precipitation in the weld metal accompanied by slight increase in the weight loss due to corrosion, for the similar weld joints [40]; whereas, for dissimilar welding using Incoloy 28 the increase in heat input to 5 kJ/mm led to slight decrease in the Mo microsegregation in the UMZ accompanied by improvement in the corrosion resistance of the weld joint [41]. Thus, it was recommended to keep the heat input for similar welds of SASS at 1 kJ/mm and at 5 kJ/mm for dissimilar welding.

A study has linked failure types, in the presence of mechanical loads, in weld austenitic SS joints to the mechanical and microstructural properties resulting from different welding conditions; Andres et. al [14] showed that some welding conditions promote the appearance of ferrite, formation of nonmetallic inclusions, and grain size and morphology variations of both austenite and ferrite phases. The behavior of the different structures under stress conditions varied and different causes for crack nucleation and propagation occurred to the GMAW and SMAW processes. Three different fracture modes were reported in this study at the welding joints; the FCAW process was influenced mainly by the first failure mode, while the other two had a mixture of the three different failure modes. The first failure mode was a geometrical stress concentrator at the weld root, due to a variation of the cross-section area of the weld that acted as the nucleation site of the crack, and further propagated through the weld or the HAZ. The second failure mode was generated at the HAZ, where the crack nucleated due to a variation in the grain size produced by the welding process, and further propagated through the edge or the center of the weld. The third failure mode appeared due to the presence of exogenous inclusions or slag generated by the welding process, which acted as stress raisers in the weld and produced the initiation and further propagation of the crack.

3.3 ROLE OF MICROSTRUCTURE ON SCC FAILURES OF AUSTENITIC SS

Corrosion attack is usually attributed to the difference in the composition and microstructures of the weld and parent metals caused by the high temperature and cooling rate; these in turn are being affected by the welding procedure used [42]. The welding defects which cause corrosion include poor adhesion, slag inclusions, crevice formation, macro and microfissures, liquid embrittlement, galvanic coupling, etc. Any of these defects can act as initiation points for corrosion attack on weld [26]. Under conditions of high chloride concentration and stagnancy, pits are developed at the vicinity of weld zone forming crevices which act subsequently as initiation centers for corrosion. It was reported that autogeneous welding has a detrimental effect on pitting resistance, as the pitting potential and critical pitting temperatures were lower for welded than unwelded steel [43]. The role of nonmetallic inclusions and left-over slag should be also emphasized. For deeper understanding of pitting corrosion, the reader may be referred to Frankel [44].

Almost, all austenitic SS grades, especially types 304 and 316, are susceptible to SCC. Other localized forms such as pitting and crevice corrosion probably precedes SCC and may act as origins where the cracks can nucleate and then propagate into the metal. These stresses may be originating from external sources (applied) or internal sources (residual stresses). Nearly, all metals are susceptible to SCC in the presence of tensile stresses in specific environments. There are three necessary conditions for SCC to occur; one factor is susceptible microstructure, the second is stress conditions either resulting from intrinsic or extrinsic sources and the third is the presence of a corrodant environment. Complete lists of environments that may cause SCC in many common metals are available in the literature [45].

In many austenitic SSs, the heat of welding causes sensitization (depletion of chromium from the matrix due to the precipitation of chromium carbides along the grain boundaries) [46]. Stress-corrosion cracks are reported to extend in the chromium-depleted regions. This has been explained in view of Cr being the element responsible for giving SS its corrosion resistance [45]. Other microstructural changes associated with sensitization are sigma phase precipitation and grain growth. Sensitization promotes intergranular corrosion [47].

SCC has been found to be related to microstructure. A study [48] showed that SCC happens in almost all microstructures: austenitic, martensitic, and ferritic-pearlitic, in discerningly order, where each microstructure produced a unique shape for SCC. The initiation point was pitting type cracks for austenite, corroded pearlite for ferrite-pearlite and corroded carbides for martensite structures. In all cases, propagation was intergranular, associated with grain boundary thickening in austenite. A recent approach has been handled in literature [49] where correlations are made between intergranular stress-corrosion cracking (IGSCC) resistance and grain boundary character and crystallographic texture for pipeline steels, however, no work could be found related to SSs. These studies have linked SSC susceptibility to crystallographic texture and grain boundary features resulting from the recrystallization and thermomechanical effects during rolling of these steels.

A failure investigation on secondary super-heater tubes in a power boiler indicated that SCC failure started at a crack originating from a highly stressed area and progressed outwards, with presence of branched cracks and pits, following an intergranular microstructural path associated with other characteristics of an IGSCC mechanism [30]. The macroscopic examination of the external and internal surfaces of in-service failed 304H tubes of a boiler (of the composition shown in Table 15.6) that exhibited SCC during service in a boiler has revealed that the tubes have experienced the occurrence of thin cracking originating at the internal surface of the tube and progressing outward circumferentially, as shown in Figure 15.7, showing the progression of the

Table 15.6 Chemical Composition of Studied Material (wt.%)

C	Cr	Ni	Mn	Si	S	P	Al	Ti	Fe
0.089	18.05	9.48	1.66	0.455	0.005	0.025	0.00359	0.00402	Bal

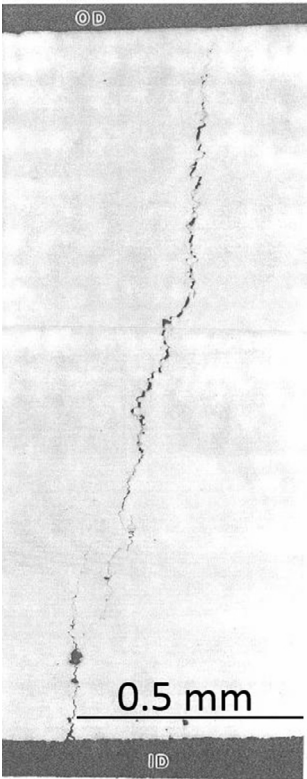
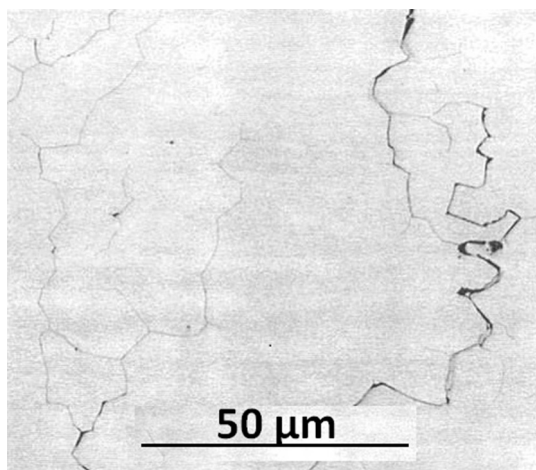


FIGURE 15.7

Full section crack [30].

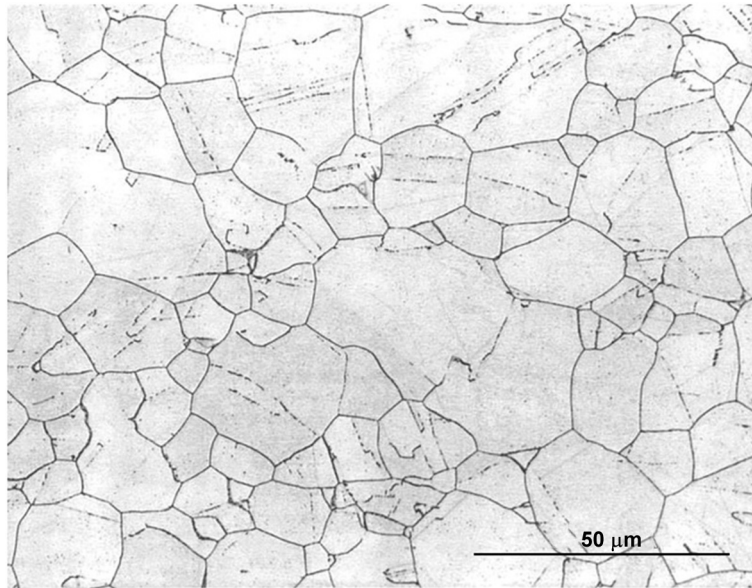
**FIGURE 15.8**

Photomicrograph showing branching near mid-wall [100×] [30].

crack from the internal surface from which it is revealed that in addition to the fine branched cracks, pitting also occurred as could be observed in association within central portions of stains. The microscopic investigations showed cracks propagating nearly through the wall as shown in Figure 15.8. The principle crack exhibited some branching near the inner surface (Figure 15.8). The cracks propagated for a distance of less than one-quarter of the tube wall and exhibited some branching. Results of microstructural examination (Figure 15.9) revealed austenitic grain structure with continuous precipitation of chromium carbides along grain boundaries, where the hardness reached 78-79 HRB at locations of continuous precipitation of chromium carbides along the grain boundaries. Figure 15.10 reveals the cracking around the grain boundaries. The recorded observations are consistent with (IGSCC) [50]. The study also showed that local heating resulting from bend solution treatment or heat input during welding of SS parts create the necessary conditions developing susceptible microstructures to SCC, the most important being the sensitization effect resulting from the precipitation of chromium carbides along the grain boundaries. Stress-corrosion cracks originate at regions of high stresses and propagate outward following the Cr-depleted zones (passes) to less stressed parts till they stop. Pitting and branching of finer cracks was associated with Cr-depleted zones. The study also showed a good correlation between SCC behavior and sensitization effects in 304H SS.

3.4 FAILURE OF A DSS AIR COOLING TUBE PLATE BY SULFIDE STRESS-CORROSION CRACKING

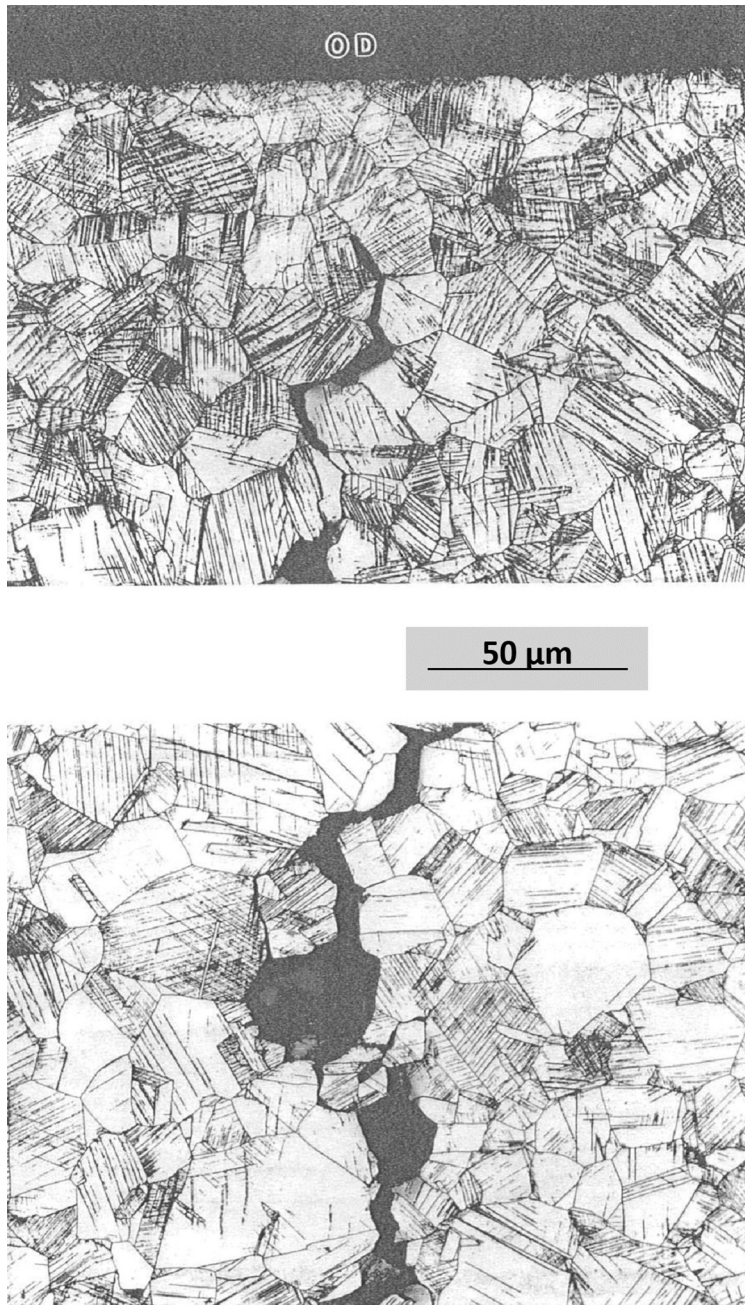
A failure case of air cooling tube-plate connections was reported in Slovakia in 2013 [51]. The failure happened after 3 weeks of service, following a welding maintenance stop. The air cooler consists of a plug header with dissimilar welded joints; the

**FIGURE 15.9**

Microstructure exhibits continuous carbide precipitation at grain boundaries [200×] [30].

internal part of the plug header is made of the 15Mo3 steel, while the tubes Ø 25 mm were made of duplex steel DIN 1.4462 welded to the tube-plate ($h = 40$ mm) (see the scheme of the header in [Figure 15.11a and b](#)). The welds were fabricated by TIG process using a special automatic welding machine with ER 309Mo wire Ø 0.6 mm, which is generally recommended for welding the given combination of steels. The chemical composition of the materials used is given in [Table 15.7](#). The working medium for this air cooler consisted of hydrogen (up to 70%), hydrocarbons (up to 28%), and hydrogen sulfide, as well as water and other admixtures. The working pressure was 13.48 MPa and working temperature 50–122 °C. The metallurgical analysis of the case by macrostructure, metallographic, microhardness, and microanalysis ([Figure 15.12](#)) showed a defected welded joint. The weld metal hardness attained values of HV 435, the chemical microanalysis (EDX) of the weld metal revealed its considerable dilution by the tube-plate material: 10.03 wt.% Cr, 6.42 wt.% Ni, 1.55 wt.% Mo.

The analysis of the collected data revealed that failure occurred as a result of single pass welding as a consequence of very small thickness of the tube-plate [51]. The failure cause was identified to be sulfide stress-corrosion cracking (SSCC). From the viewpoint of SSCC, the hardness should not exceed the values of HV 248 according to NACE standard for low-alloy steels (for duplex steels the hardness limit is higher, namely, HV 285). The dilution of the weld metal to the previous composition was suitable for martensitic transformation upon fast cooling of welding process, as show

**FIGURE 15.10**

Microstructure adjacent to through-wall crack, etched electrolytically using oxalic acid solution [200×] [30].

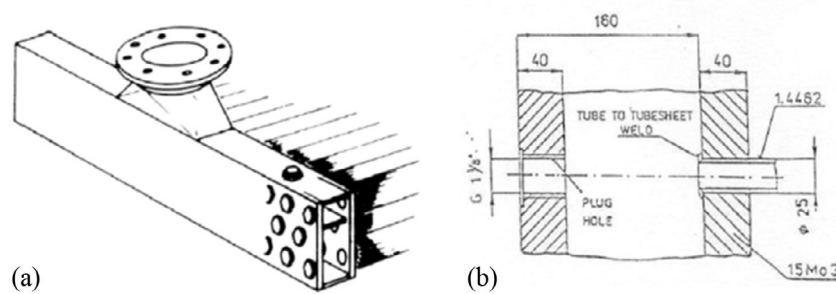


FIGURE 15.11

A schematic of the plug header in the aircooler [51].

Table 15.7 Chemical Composition of Studied Material (wt.%) [51]

Material	C	Mn	Si	S	P	Cr	Ni	Mo	Nb	N
15Mo3	0.130	0.77	0.22	0.10	0.009	-	-	0.12	-	-
1.4462	0.025	1.69	0.31	0.003	0.033	22	5.45	3.04	0.53	0.135
ER306Mo	0.34	1.49	0.40	0.012	-	24	14.2	2.74	0.13	-

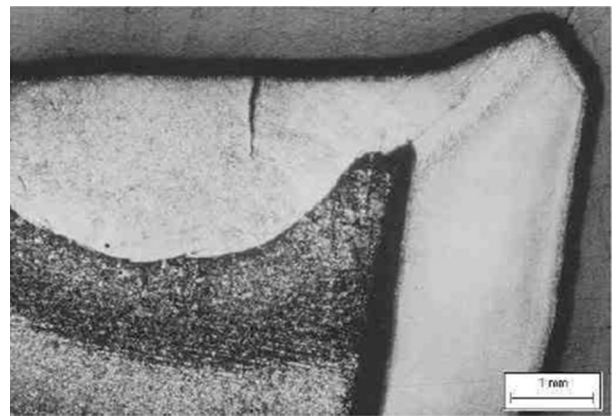


FIGURE 15.12

Macrosection of defected welded joint [51].

in Figure 15.13. This has provided the conditions required to form SSCC ((1) suitable microstructure (martensite+high hardness), (2) high percentage of sulfide compounds, and (3) stresses (residual stresses from welding process)). The combination of these factors (existing after welding of the previous components) led to the failure in the tube-plate by SSCC, as shown in Figures 15.14 and 15.15.

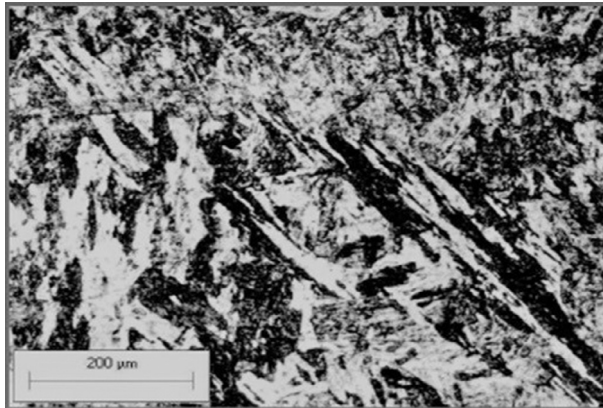


FIGURE 15.13

Martensitic microstructure of the WM [51].

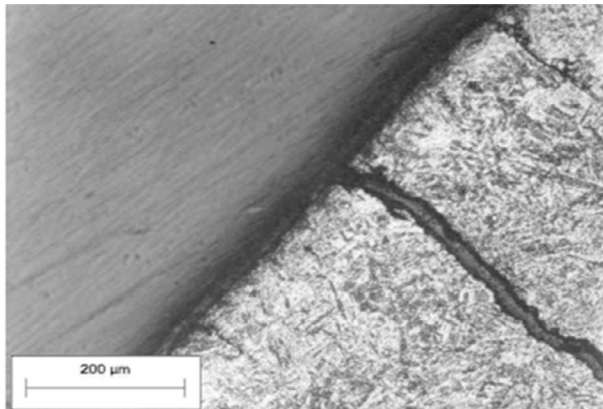


FIGURE 15.14

SSCC in the weld metal [51].

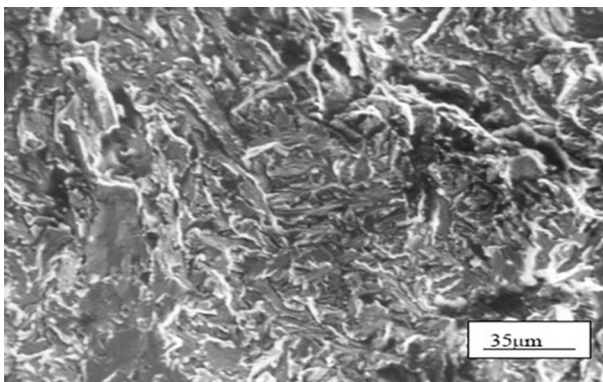


FIGURE 15.15

SEM of SSCC in the weld metal [51].

4 CONCLUSION

It should be kept in the mind of design and material engineers that the service conditions for power boilers provides the three necessary conditions (susceptible microstructure, state of stress, and corroding environment) causing SCC to end the life of parts. If stresses are absent, other forms of corrosion (pitting) may appear. During installation and assembly, it should be remembered that welding of attachments provides two out of three parameters essential for SCC existence. The wrong selection of the material or welding process may lead to design or metallurgical conditions promoting SCC.

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Degradation of protective PVD coatings

16

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1 INTRODUCTION

Nowadays, coatings obtained by physical vapor deposition (PVD) have been widely used to improve performance of construction or mechanical elements, extend service life of tools, especially high-speed cutting tools due to their excellent tribological properties and high hardness, and increase service life of blades in gas turbines due to their high wear and oxidation resistance.

In previous review papers [1–6], PVD coatings were discussed mainly in aspects in using on cutting tools (Table 16.1). Also, very deep studies about crystallographic structure were performed [3]. However, there was lack of studies about deformation mechanisms, fatigue, and erosion resistance of PVD coatings.

Due to the size of grains in PVD coatings (in most cases below 100 nm), PVD coatings are classified as nanostructured materials [1,7]. Properties of PVD coatings are mainly influenced by their chemical composition, structure (grain size and shape, defects density, coating density), and residual stresses that depend on method and parameters of deposition. PVD coatings possess compressive stresses that influence their properties: high hardness, high elastic modulus, and high fracture strength [1,8–20]. They also have high oxidation resistance and low friction coefficient [21,22].

Table 16.1 Summary Previous Reviews

Previous Reviews	Goals
Inspektor and Salvador [2]	Mono- and multi-layer nanocomposite Ti- or Al-based coatings for cutting tools; dependence of hardness on annealing temperature; hot hardness; thermal stability; resistance to flank wear
Andrievski [1]	Mono- and multi-layer Ti-based coatings; hardness, texture, deformation under indentation; grain-size effect on hardness, high-resolution TEM study; an effect of the magnetic field on the predominant texture
Mayrhofer et al. [3]	Hard PVD coatings for wear-resistant applications; study an influence of some deposition parameters on coating microstructure including high-resolution TEM investigation (dislocation generation and movement, structure of grain boundary) and some relation between microstructure and mechanical property (inverse Hall-Petch effect); microstructural (lattice and crystallographic structure) evolution of hard ceramic coatings during a post-deposition annealing; diffusion processes
Navinsek et al. [4]	PVD coatings were compared with electroless coatings and electrodeposited coatings; analyzed corrosion property, coatings porosity; PVD coatings—substitute for some galvanic coatings
Sproul [5]	Descriptions of PVD techniques; PVD coatings for tools—TiN, TiAlN, TiCN ZrN, TiZrN, CN _x , Al ₂ O ₃ coatings; superlattice coatings
Van Stappen et al. [6]	PVD coatings (mainly TiN, Ti(CN) (TiAl)N and CrN coatings) for metal cutting applications

PVD process of coatings deposition encompasses three different techniques: arc evaporation, magnetron sputtering, and ion plating, each of which allows obtaining coatings with different mechanical and structural properties [5]. Each of the deposition parameters such as bias voltage, coil current, and temperature of substrate influences the structure and properties of PVD coatings [5,8,17,23–30]. Later in this chapter, the influence of each mentioned deposition parameter on coating structure and properties will be discussed. It should be noted that it is difficult to observe a monotonic change of a property of coating with a deposition parameter increasing continuously. Very often, one of the properties changes monotonically up to some level and afterward it stops at this level or is observed a change in opposite direction. Moreover, depending on a coating deposition technique a change of one deposition parameter may have different influence on coating property.

Plastic deformation of PVD coatings is limited by elastic and plastic deformation of the substrate, which has lower hardness than PVD coatings. The hard thin PVD coating cannot often follow plastic deformation of the substrate without failing [31–36]. Performed up to now investigations show that deformation and degradation mechanisms of PVD coatings depend not only on mechanical properties of the coating and substrate but also on the method of degradation.

The aim of this review is to present the influence of deposition parameters on PVD coating properties, the influence of PVD coatings deposition on the strength of coating-substrate system, and to present deformation mechanisms of mono- and multi-layers PVD coating. Moreover, there is presented an influence of PVD coatings deposition on fatigue strength and degradation caused by erosion.

2 INFLUENCE OF DEPOSITION PARAMETERS ON PVD COATING PROPERTIES

Deposition parameters (substrate temperature, gas pressure, bias voltage, argon and nitrogen flow rate, ion energy) influence structural (crystalline size, lattice parameter, number and size of defects, grain size, coating density, coating morphology) and mechanical (hardness, adhesion, fracture toughness) properties. Some relations between deposition parameters and coating properties are presented in Figure 16.1. An influence of each deposition parameter is discussed in detail below.

Deposition parameters, especially substrate temperature and sputter pressure, have an influence on the growth of thin coatings (grain initiation and growth). The most known model that shows the coating structure is the Thornton's model (Figure 16.2), despite the fact that some modifications of this model have been made [37,38].

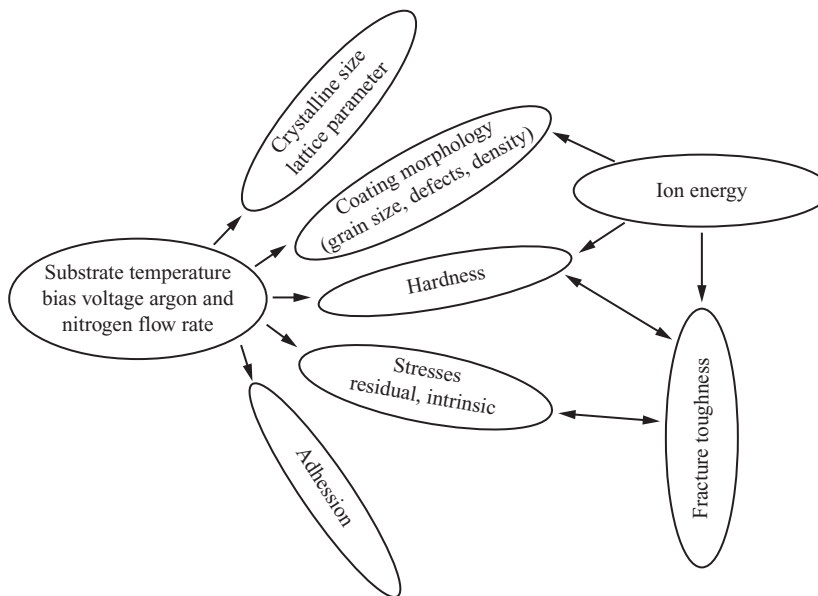
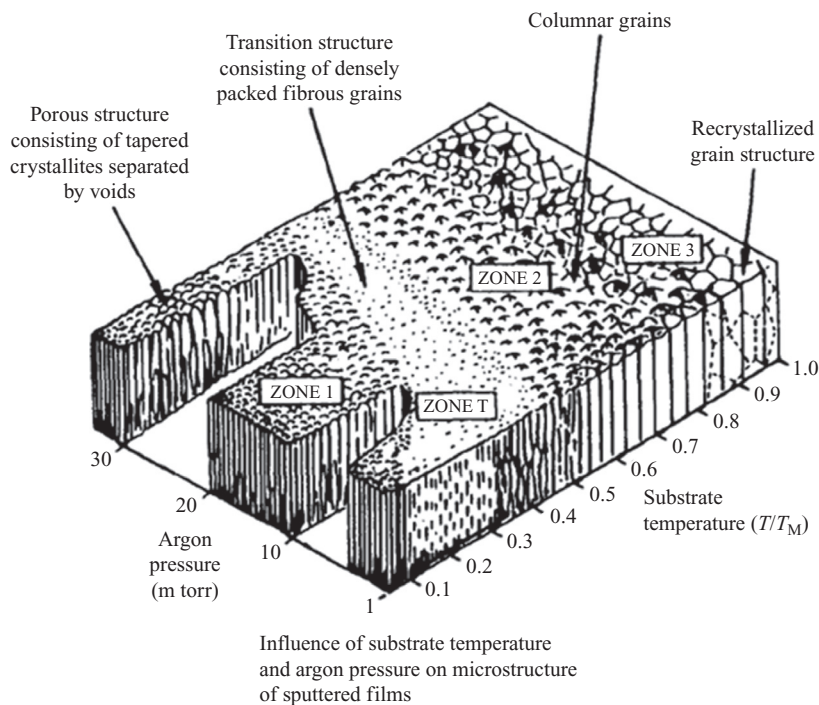


FIGURE 16.1

Influence of deposition parameters on PVD coating properties.

**FIGURE 16.2**

Thornton's model of thin coating structure.

In the Thornton's model exists four zones, in which different structures are formed. At low substrate temperature below $0.1 T_m$ (zone 1), the nucleation density is high and the coating growth proceeds in vertical direction. In this zone, fiber-like porous structure develops. The diameter of fibers is in the range between 1 and 10 nm; with increasing T_s/T_m , the fiber diameter increases. At higher temperature ($0.2 < T_s/T_m < 0.4$) exists a zone *T*, in which coating has an inhomogeneous structure along its thickness. At the substrate, fine crystallites grow, and next cone-shaped columnar grains, while in the upper part of thick coating are columnar grains. Further increase of substrate temperature ($T_s/T_m > 0.4$), causes the coating to have a homogeneous structure that is composed of grains in the form of columns, whose height is as same as the coating thickness. The grain boundaries are nearly perpendicular to the coating plane. In zone 3, the structure is characterized by equiaxed three-dimensional grains.

Most PVD coatings are deposited under conditions that allow obtaining coatings with columnar grains oriented approximately parallel to the direction of growth. However, not all columnar grains extend the coating thickness, indicating that some grains are nucleated during the coating growth. This indicates that structures described in the Thornton's model are only present in an ideal coating. In real

coatings are present some impurities and defects that limit the grains growth. The most common impurity is oxygen, while micro-droplets are the most common defects, particularly in coatings deposited by arc evaporation and by magnetron reactive sputtering. The number and size of micro-droplets depend on deposition parameters like bias voltage [17], substrate temperature [39], argon and nitrogen flow rate [11], and gas pressure [40].

The investigations of an influence of *negative bias voltage* U_B on mechanical and structural properties of produced coatings showed a great importance of a deposition method. In case of an arc evaporation method, a decrease of bias voltage to -100 V caused the decrease of grain size in a CrN coating, while in the case of reactive magnetron sputtering method, a slight increase of grain size was noted [17,24]. Further decrease of bias voltage caused a slight increase of grain size in case of CrN arc-evaporated coating, while in the CrN magnetron-sputtered coating—a significant decrease of a grain size.

A decrease of bias voltage caused an increase of lattice parameter [8,25]. In the case of arc evaporation method, the lattice parameter in TiAlN coatings increased from 0.4173 to 0.4231 nm (1.4%) with a decrease of negative bias voltage from -40 to -140 V [8], while in the case of CrN_x coatings deposited by reactive unbalanced magnetron sputtering method, the lattice parameter increased from 0.4117 to 0.4225 nm (2.6%) with a decrease of negative bias voltage from 0 to -200 V [25]. The structure of CrN_x coatings changed from open dendritic film grown at low bias voltages to dense structures at higher ones. Moreover, in arc-evaporated coatings, with decreasing negative bias voltage decrease a coating roughness, a size of defects and a defect density [17].

Bias voltage has an influence also on mass density [25,27], residual stresses [8,24,27,41,42], adhesion [17,42], and hardness [8,25,29,41,42] of coating. In non-hydrogenated a-C films deposited by DC magnetron sputtering, a decrease of bias voltage caused an increase of mass density up to some saturation level that reached at approximately -100 V [27]. This saturation level depends on argon pressure; with increasing argon pressure the saturation level decreases and, as a consequence, the mass density also decreases.

The decrease of bias voltage caused a decrease of intrinsic stresses in magnetron-sputtered coatings (CrN and TiN) [24,41]. Investigations of an influence of bias voltage decrease on intrinsic stresses in arc-evaporated coatings gave two opposite results. According to Ahlgren and Blomqvist [8], in TiAlN coating the intrinsic stresses decreased from -1.7 GPa for $U_B = -40$ V to -5 GPa for $U_B = -200$ V (increased the compressive stress), while according to Grant et al. [42] and Karlsson et al. [43], the decrease of substrate bias voltage caused an increase of intrinsic stresses in CrN coatings (from -6 GPa for $U_B = -50$ V to 2 GPa for $U_B = -200$ V) and in TiN and TiC_xN_{1-x} coatings (for TiN coating, stresses increase from -6 GPa for $U_B = -50$ V to -1 GPa for $U_B = -700$ V); stresses were changed from compressive to tensile stresses. However, in case of TiC_xN_{1-x} coatings, first intrinsic stresses decreased, but later they increased. The bias voltage at which intrinsic stresses began to increase depending on the carbon content in a

TiC_xN_{1-x} coating. With increasing carbon in coating (higher x), intrinsic stresses began to increase at lower bias voltage.

Investigations of an effect of bias voltage on coating hardness showed that the decrease of bias voltage first caused an increase of coating hardness, but later with further increase of bias voltage the coating hardness decreased regardless of the deposition method, for example, in coatings deposited by arc evaporation [8], unbalanced magnetron sputtering technique [25,41], and ion plating [29]. However, the rate of decrease of the hardness was dependent on the deposition method.

An increase of bias voltage magnitude (from 0 to -300 V) caused an increase of coating adhesion in CrN_x coatings prepared by reactive magnetron sputtering [21], while in arc-evaporated films (CrN) caused a decrease of coating adhesion [17,44].

Summarizing, bias voltage affects many coating properties and is one of the most important parameters of coating deposition.

Temperature of a substrate, similar to bias voltage, is a very important deposition parameter. According to Thornton's model, substrate temperature has an influence on coating morphology. However, many investigations [24,26,29,39,45] showed that substrate temperature has also an influence on a coating grain size and hardness.

Substrate temperature influences thermal stresses, which can be calculated from the well-known equation:

$$\sigma_{th} = \frac{E_f}{1 - \nu_f} (\alpha_f - \alpha_{sub}) (T_{dep} - T_{room}) \quad (16.1)$$

where E_f is the elastic modulus of coating, ν_f is the Poisson's ratio of coating, α_f is the thermal expansion coefficient of coating; α_{sub} is the thermal expansion coefficient of substrate; T_{dep} is the depositing temperature, and T_{room} is the room temperature.

From Equation 16.1, it can be seen that thermal stresses depend on substrate temperature and thermal properties of the coating and the substrate, which can be either tensile or compressive. Taking into account that in all PVD methods, the deposition temperature is higher than room temperature, and thermal expansion coefficient of coating is lower than the thermal expansion coefficient of substrate, then the thermal stresses are compressive. In general, compressive stresses in coating are advantageous because they improve wear resistance. However, very high compressive stress in PVD coating, similarly to tensile stress, may lead to the decrease of adhesion and to delamination. Experimental investigations [29] showed that despite compressive value of thermally induced stresses, intrinsic stresses of Cr-N coating increased toward a tensile mode with increasing substrate temperature, but the rate of increase of the intrinsic stresses was dependent on the deposition method: vacuum arc evaporation caused more intensive increase of intrinsic stresses than ion plating.

A substrate temperature also influences coating hardness: in the case of the CrN coating [24,46] the hardness increased with increasing substrate temperature, while in case of TiN coatings, the hardness decreased [39,45]. The substrate temperature has also an effect on the adhesion. In the case of TiN arc-evaporated coatings, the adhesion increased with increasing substrate temperature, while in the case of CrN coatings, the decrease in adhesion occurred [39]. However, Krella

and Czyżniewski's investigations of TiN and CrN arc-evaporated coatings did not confirm these relations [45,46].

With increasing substrate temperature, the microstructure of the Cr-N coatings is changing from a columnar form to a "more dense and uniform" structure, and the surface becomes more and more smooth [26,45]. On the other hand, in Gahlin et al. [39], the opposite results were obtained: the increase of substrate temperature caused an increase of roughness of TiN and CrN arc-evaporated coatings and an increase of metal drop density. In ion plating and vacuum arc evaporation, with increasing substrate temperature the size of grains increased [29].

Ion energy in ion-beam assisted magnetron sputtering had influence on the grain size, microstrain, hardness, and fracture toughness [15,19]. In the case of the CrN coating, an increase of ion energy up to 800 eV decreased a grain size and hardness, and increased fracture toughness, but further increase of ion energy caused an opposite effect [19]. However, microstrain increased continuously with an increase of ion energy, but a rate of microstrain increase slowed down with an increase of this deposition parameter.

In nitrogen-based coatings, *a rate of nitrogen flow* and the ratio of Ar/N₂ during the coating deposition had an influence on coating structure, hardness, and adhesion [9,11,30]. In reactive magnetron sputtering method, an increase of nitrogen flow rate caused nearly linear increase of nitrogen amounts incorporated in Cr-N coating [30], and also increased crystalline size and lattice parameter [9]. With an increase of nitrogen flow from 20 to 60 sccm, the coating hardness in the Cr-N coating increased from 26.9 to 44.9 GPa [9]. This increase was likely caused by an increase of compressive residual stress due to the insertion of nitrogen atoms in the CrN structure and an increase of lattice parameter from 4.2068 to 4.2566 Å. However, further increase of nitrogen caused a decrease of coating hardness [30]. In opposition to the hardness, with the increase of content of nitrogen in these Cr-N coatings the critical load, which is often treated as a measure of adhesion, decreased [30]. On the other hand, the hardness of Cr-based coatings decreases with increasing argon and nitrogen flow rate at a constant Ar/N₂ ratio [11].

Not only deposition parameters have influence on PVD coating properties (Table 16.2). An increase of *coating thickness* causes a change of a preferred growth orientation [9,55] and microstructure [59], and has an influence on residual stress and coating hardness [9,33,55,57,58]. An experimental data show opposite results. According to Barata et al. [9], an increase of thickness of CrN coating from 2.5 to 4.2 µm deposited by reactive r.f. magnetron sputtering caused a decrease of coating hardness from 44.9 to 27.1 GPa, while an increase of thickness from 2.5 to 5 µm of TiN, TiAlN, and TiCN coatings deposited by arc ion plating caused an increase of coating hardness from 700 to 3000 HV [58]. Also, an increase of thickness from 4 to 12 µm of TiN coating deposited by cathodic arc evaporation caused an increase of coating hardness from 21.7 to 27.4 GPa [33]. Chou et al. [55] have shown that as thickness of TiN coating increased from 0.32 to 0.75 µm the most intensive increase of coating hardness was noted, further increase of coating thickness had a slightly influence on coating hardness. Moreover, the increase of TiN coating

Table 16.2 Factors, Apart from the Deposition Parameters, Affecting Coating Properties

Factor	Effect	References
Chemical composition of coatings (e.g., addition Al to CrN or TiN coatings)	• Coating hardness • Residual stress • Toughness • Stiffness • Adhesion • Erosion resistance • Friction coefficient	Batista et al. [47], Batista et al. [48], Brizuela et al. [11], Ding and Zeng [49], Jianxin et al. [50], Uchida et al. [51], and Wang et al. [52]
Substrate hardness	• Coating hardness • Residual stress • Erosion resistance	Bhowmick et al. [10], Korunsky et al. [53], Krella and Czyżniewski [54], and Krella [33]
Coating thickness	• Grain preferred orientation • Microstructure • Coating hardness • Residual stress stiffness • Adhesion • Erosion resistance • Friction coefficient	Barata et al. [9], Chou et al. [55], Immarigeon et al. [56], Jeon et al. [57], Kim et al. [58], Korunsky et al. [53], Krella [33], Lang and Yu [14], Ma et al. [59], and Scheerer et al. [60]

thickness caused an increase of packing factor and texture coefficient, and a decrease of residual stresses [55], and also a decrease of a crystalline size and a lattice parameter [9].

The strength of PVD coating depends also on *substrate properties*: substrate hardness, elastic modulus, and thermal properties [10,41]. Bhowmick et al. [10] have shown that the substrate properties influence the magnitude of the residual stresses in the coating, and also the lattice parameter of the thin PVD coating. With an increase of a substrate hardness, the lattice parameter, and residual stress increases. The increase of residual stress in the coating was proportional to an increase of lattice parameter. On the other hand, taking into account that PVD coatings are typically deposited at temperatures in the range 400–450 °C, the thermal mismatch between substrate and coating introduces thermal stresses (see Equation 16.1). If thermal expansion coefficient of a substrate is higher than that of a coating, then negative (compressive) thermal stress are generated in the coating. If coefficient of thermal expansion of a substrate is lower than that of a coating, it introduces positive (tensile) thermal stress in the coating, and eventually tensile stress in the coating is generated. Tensile stresses decrease the coating strength and wear resistance [41].

An *addition of aluminum* to CrN or TiN coatings deposited by magnetron sputtering, cathode arc evaporation and by electron beam evaporation technique caused an increase of coating hardness, elastic modulus, adhesion and erosion resistance, and a decrease of friction coefficient [11,47–51].

Properties of coatings with addition of Al, similarly to coating without Al, depend on deposition parameters. In case of CrAlN coatings, their properties depend on Al/Cr ratio: with an increase of Al/Cr ratio up to 0.4-0.5, hardness and elastic modulus increased, but further increase of Al/Cr ratio caused a decrease of these properties [49]. A decrease of bias voltage magnitude from -50 to -260 V caused an increase of coating density, with lowered coating defects, increased coating hardness, elastic modulus, and toughness, and decreased the grain size in chromium aluminum nitride coatings deposited via magnetron sputtering [31]. However, investigations shown in Wang et al. [52] revealed the decrease of hardness, elastic modulus, and critical load in scratch test, as a result of addition of aluminum to CrN coating deposited by electron beam evaporation technique.

3 DEFORMATION OF MONOLAYER COATINGS

The deformation mechanism of PVD coatings is influenced by the structural morphology of the coating and the hardness of substrate, and is preceded by the elasto-plastic deformation of the substrate [10,61]. High-resolution TEM study of TiN coating revealed the presence of edge dislocations inside columnar crystallites [1,3]. The dislocations confirmed possibility of plastic deformations in hard brittle TiN coating. Nevertheless, plastic deformation of TiN coating is very small, and high tensile stress caused by bending, stretching, or heating leads to intergranular fracture (Figure 16.3) [41].

Investigations of an influence of substrate finishing on the deformation and failure mechanisms showed that systems with microcracked surface of the substrate

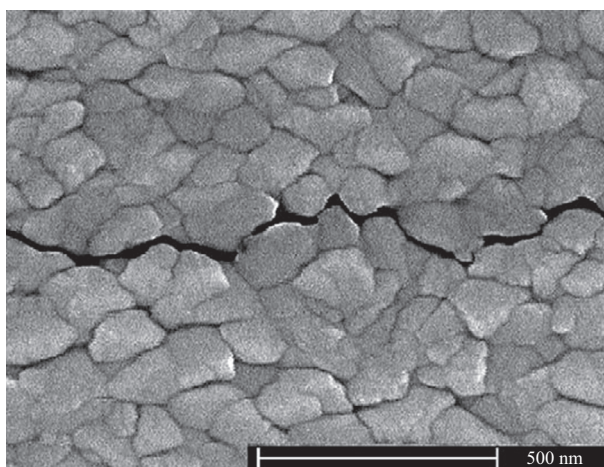


FIGURE 16.3

Intergranular fracture caused by tensile thermal stresses.

Bielawski and Seo [41], permission from Elsevier.

tended to exhibit small chipped or spalled regions, while on substrates whose surface is almost free of microcracks, adhesive failure is evidenced as a complete detachment of the coating at relatively high loads, usually initiated at through-thickness cracks (cohesive failure) [62].

The initial deformation of the coating, at very low indentation depth, was found to be elastic or elastic-plastic [10]. At higher indentation depths, when a substrate starts to flow plastically, a PVD coating is likely to follow the deformed substrate. Based on experimental observations, Bhowmick et al. [10] have proposed a model of PVD coating deformation (Figure 16.4a). According to this model, deformation of columnar-grained PVD coating occurs via shearing and sliding at intercolumnar boundary under the action of a shear stress. This model was later confirmed by Jayaram et al. [32] in testing thin TiN coatings (Figure 16.4b). When the plastic

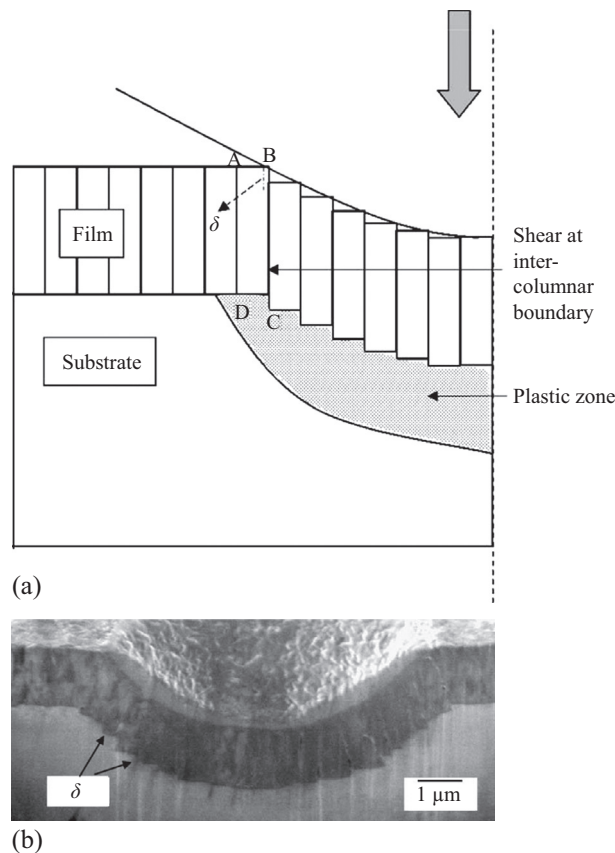
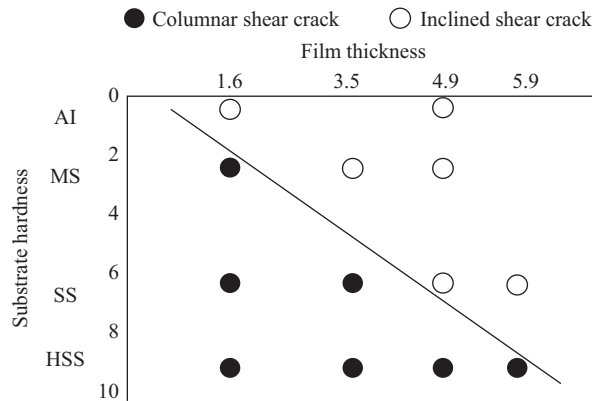


FIGURE 16.4

Model of deformation of columnar-grained coating [10] (a) and cross-section of an indented TiN columnar coating [32] (b), permission from Elsevier.

**FIGURE 16.5**

A plot showing the change in failure modes with film thickness and substrate hardness.

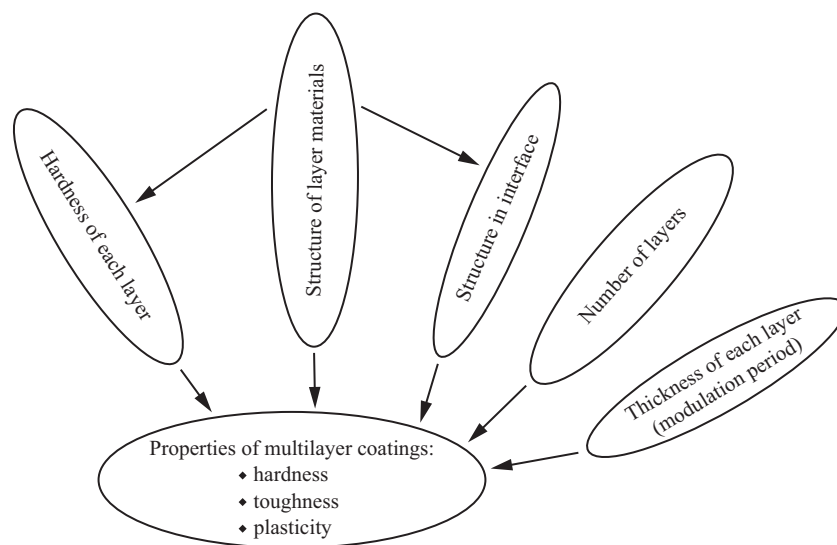
Jayaram et al. [32], permission from Elsevier.

deformation in the substrate rises to a level that the coating is not able to deform elastic-plastically and absorb the deformation energy, then it breaks causing the relaxation of strain energy. According to Figure 16.4a, columnar-grained coatings should dominate columnar shear cracks. However, investigation performed by Jayaram et al. [32] showed that besides columnar shear cracks occur inclined cracks and the kind of cracks generated in columnar TiN coatings depends on the coating thickness and substrate hardness (Figure 16.5). Columnar shear cracks arise mostly in thin films deposited on hard substrates, while in thick coatings deposited on soft substrates arise inclined shear cracks.

The substrate hardness has an influence on stress at which plastic flow of substrate begins, while with increasing coating thickness increases the number of defects in coating structure. In the case of a soft substrate, the deformation of a coating develops due to sliding and spitting of the adjacent columns, and circumferential cracks arise at the edge of a spherical indentation [1,47,61,63]. In the case of a hard substrate, the deformation of a coating develops due to shear at intercolumnar boundary and radial cracks are formed, but circumferential cracks hardly arise [10,61]. In case of dense noncolumnar coatings, brittle circumferential cracks arise despite the hardness of a substrate [61].

4 PROPERTIES OF MULTI-LAYER COATINGS

Multi-layer coatings consist mostly of several repetition of two layers—one harder and one softer. They may be made of ceramic and metallic layers, or two different ceramic layers. Properties of layer material influence properties of multi-layer coating. Schematic correlation between properties of layers and properties of multi-layer coating is presented in Figure 16.6.

**FIGURE 16.6**

Correlations between properties of layers and properties of multi-layer coating.

Multi-layer coatings have better wear resistance and resistance to fracture (higher critical load) than monolayer coatings [64], therefore, increasing the interest in them. Hardness, elastic modulus, and plasticity of multi-layer coatings are between relevant properties of each layer [65–72]. Properties of multi-layer coating also depend on the number of layers and the thickness of each layer. The number of layers influences the number of interfaces that have great effect on coating properties.

According to Holleck and Schier [73], the thickness of interfaces is between 1 and 3 nm depending on the kind of materials meeting together. This leads that with increasing number of layers an interface volume increases. The interface volume reaches ~2–6% for a 100 layer coating, up to 60% for a 1000 layer coating with thickness of 5 μm [73]. Therefore, with increasing layer number and decreasing layer thickness the interfaces play dominant role in multi-layer coating properties and deformation. The interface volume is considered as a binder phase with hardness between the hard and soft layer material (e.g., ceramic and metallic layer). The structure of the materials meeting at the interface plays a predominant role on the structure of interfaces (coherent, incoherent or semicoherent), and energy and stress distribution. Every new layer in a multi-layer arrangement tends to grow in a close structural relation to the preceding layer, so large difference in structural properties between the meeting layers influences an increase of intrinsic stress at the interfaces. SEM investigations [72,73] showed that the interfaces are the places, at which cracks are often initiated, deflected causing a change in the direction of propagation, or a reduction of crack propagation rate or blocked. Thus, it is important to select materials with similar crystallographic structure and lattice parameter to minimize the interfacial stress, and the risk of crack initiation and layer delamination. In the case

of ceramic multi-layer coatings, with increasing number of layers, the hardness of the whole coating increases [73,74].

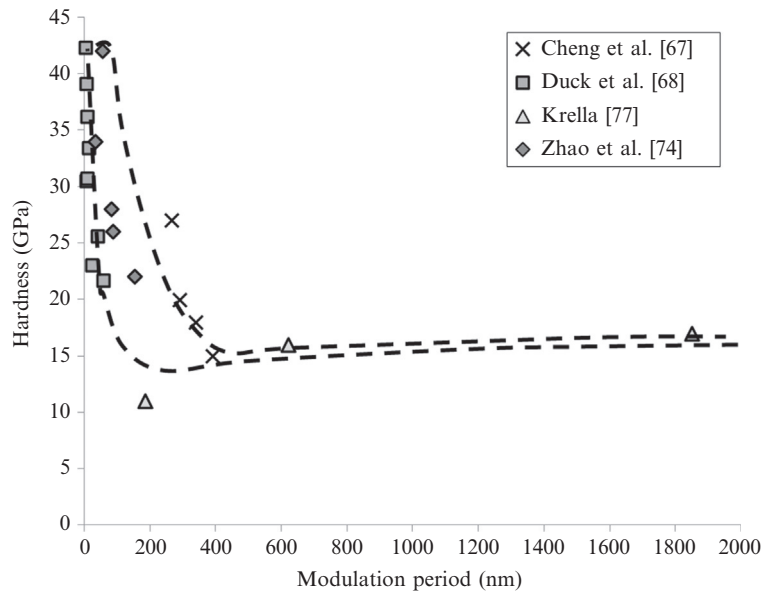
During deformation, soft layers act as buffer layers that reduce an interfacial stress and increase strain [65]. A Ti/TiN multi-layer coating with increasing thickness of soft Ti layers toward the substrate possess the better wear resistance than coatings with decreasing or stable thickness of Ti layers [65]. Bromark et al. [75] and Kim et al. [69] have noticed that with increasing the amount of Ti in Ti/TiN multi-layer coating, hardness of the whole coating decreases. A relative amount of metallic element (e.g., Ti) in a coating has been proposed as a coating parameter that correlates with the hardness of the whole multi-layer coating or with residual stresses in the coating. Similar results (the decrease of hardness with increasing relative amount of metallic element in a coating) were obtained in Cheng et al. [67] and Wieceński et al. [70]. For example, in case of Ti/TiN multi-layer coatings, when the thickness of metallic Ti layer increased from 0 to 150 nm with an unchanged thickness of the ceramic layer of 240 nm, so the relative amount of metallic element in coatings increased, the hardness decreased from 32 to 16 GPa [67]. Similar effect was obtained in Cr/CrN multi-layer coatings [70]. Based on mentioned correlation and the rule of mixture, a formula for a hardness of a multi-layer coating has been proposed [76] in the form of:

$$H_{\text{composite}} = \frac{t_1}{t_{\text{total}}} H_1 + \frac{t_2}{t_{\text{total}}} H_2 \quad (16.2)$$

where t is thickness, H is hardness, 1 and 2 mean compounds of each layer, for example, Cr and CrN in Cr/CrN multi-layer coating. However, this formula and the parameter of relative amount of metallic element in a multi-layer coating are not suited for data from Duck et al. [68], Krella [77], and Zhao et al. [74].

In Bull and Jones [78], Duck et al. [68], and Zhao et al. [74] was used a modulation period, which is a spacing of the repeat in the structure, as a parameter that characterizes a multi-layer coating. With increasing a modulation period, the hardness of the multi-layer coating decreases. However, both modulation period and relative amount of softer compound are in opposition to the results presented in Krella [77] and Maurer and Schulz [79]. On the other hand, the thicknesses of soft Ti layer in Krella [77] and Maurer and Schulz [79] were in the range of 90 to 925 nm, while in other works, they were in the range of from 10 to 150 nm. Thus, the problem in the lack of consistency is probably connected with the layer thickness and interface volume, which is much lower in case of Krella [77] and Maurer and Schulz [79] than in other coatings. Taking into account all mentioned data and modulation period one can obtain a relation shown in Figure 16.7. Figure 16.7 shows that if multi-layer period is of micrometer sized, the increase of content of soft Ti has no effect on the hardness of the whole multi-layer coating. On the other hand, if modulation period decreases to nanometer size, the hardness of multi-layer coating increases.

The thickness of soft metal layers has also an influence on stress intensity factors for macroscopic crack propagation: with increasing metal layer thickness, the stress intensity factor and the critical stress intensity for crack reinitiation increase as well [80,81].

**FIGURE 16.7**

Relation between hardness and modulation period.

Zhao et al. [72] has obtained that in a multi-layer coating are generated lower maximum radial stress, shear stress and equivalent stress than in a monolayer coating under the same load. This shows that plasticity of multi-layer coating is higher than that of the monolayer. On the other hand, in tribological test (ball-on-disk tests), Ti/TiN multi-layer coatings had worse wear resistance than a TiN monolayer coating [67]. According to Cheng et al. [67], with increasing thickness of Ti layer, plastic deformation of this metallic layer was easier and increases the mismatch of plastic deformation between Ti and TiN layers. Thus, if Ti layer is quite thick, a hard and brittle top TiN layer is removed at lower stress than in case when Ti layer is much thinner than TiN layer.

5 DEFORMATION OF MULTI-LAYER COATINGS

As it was mentioned earlier, in deformation process of multi-layer coatings, soft layers play a significant role. They act as buffer layers that favor energy dissipation by shear deformation, the drop of intrinsic stress and the increase of strain [61,65,69,82]. Investigations of TiAlN/CrN multi-layer coatings showed that with increasing periodic thickness from 7 to 18 nm, increases resistance to fracture and ability to deformation with a substrate [64].

In metal-ceramic multi-layer coatings, cracks are initiated at interfaces and developed in hard and brittle ceramic layers and block at next interface [80,83,84]. Huang and Zhang [80], based on microscopic observations, showed that two different damage modes in metal-ceramic multi-layer laminates occur (Figure 16.8). According to Figure 16.8, in one mode a zone of multiple cracks is formed near the tip of a macroscopic crack, while in the second mode the continuous propagation of one of the cracks occurs in the ceramic layers with ductile metal layers remaining untouched. However, microscopic investigations of the indentation zones in Cr/CrN multi-layer coatings revealed that most of the cracks were inclined at $\sim 45^\circ$ to the surface of the coating (Figure 16.9) [84]. Cracks were not located under the tip of the indentation, but they were located at an inflection zone and formed multiple cracking in several subsequent ceramic layers. Crack growth depends on the thickness of soft layer. For thinner soft layer, cracks inclined at the interface and propagate along the interface causing local delamination and also propagated through the thickness of ceramic layer (Figure 16.9a and b). In case of thicker soft layer (Figure 16.9c), cracks propagated initially through ceramic layer.

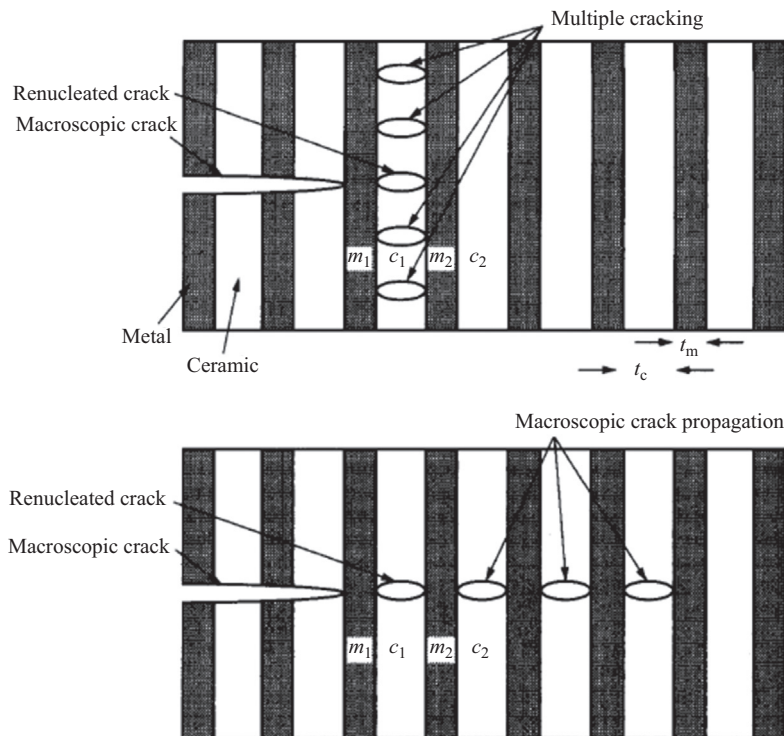
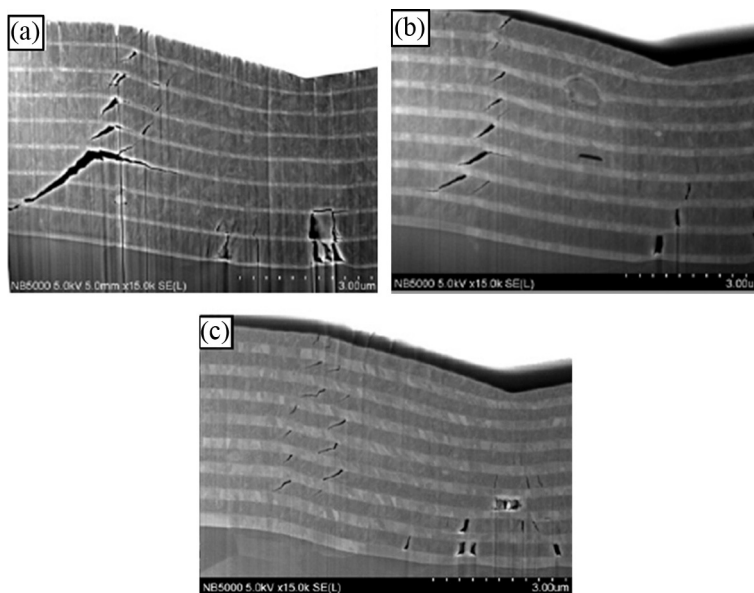


FIGURE 16.8

Two modes of laminate fracture showed in Ref. [80], permission from Elsevier.

**FIGURE 16.9**

Cross-sections of the indentation zones in Cr/CrN multi-layer coatings, load 1 N;
 (a) thickness of Cr layer: 0.09 μm , CrN layer: 0.56 μm ; (b) Cr layer: 0.17 μm , CrN layer
 0.48 μm ; (c) Cr layer: 0.24 μm , CrN layer 0.39 μm .

Wicinski et al. [84], permission from Elsevier.

Although cracks were not observed in ductile Cr layers, the plastic deformation in the form of a rotation, sliding, and twinning of the columnar Cr grain was noted [84].

In ceramic-ceramic multi-layer coatings, cracks developed via through-thickness grain boundary shearing and sliding, similar to monolayer coatings [82,85] or via intergranular shear sliding, concurrent with crack formation in the hard nanocomposite layers due to lack of deformability under compressive stress [86].

Transmission electron microscopy observations of a deformed TiN/NbN multi-layer coating revealed that shear occurs on $(1\ 0\ 1)$ and $(1\ 0\ \bar{1})$ planes inclined at $\sim 45^\circ$ to the surface of the film with dislocations moving across the multi-layer boundaries and along the columnar grain boundaries [82]. When these grain boundaries grow through the complete thickness of the film, the deformation (shearing along grain boundaries) leads to removal of a section of the coating. This removal is associated with the pores that were found along these boundaries. Next process that took place during deformation (indentation) was the rotation of the lattice planes [82]. Performed investigations confirmed that plastic deformation of ceramic multi-layer coatings is possible and the deformation mechanisms are the mechanisms that occur in nanomaterials.

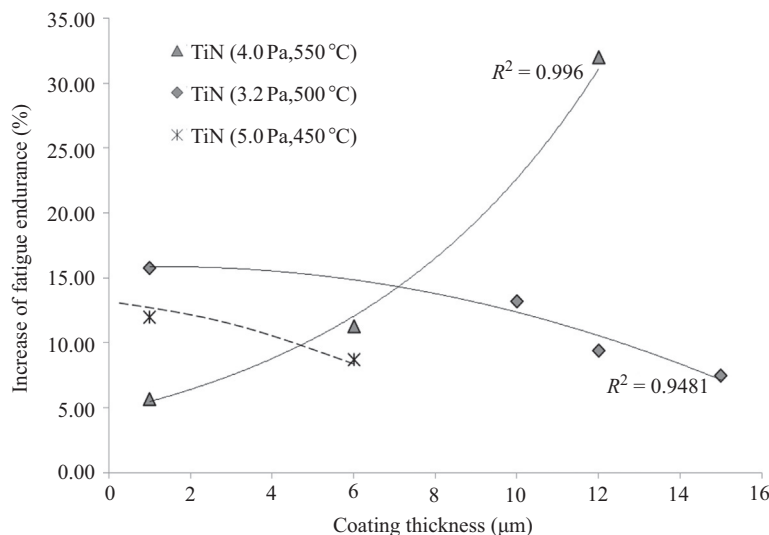
6 FATIGUE STRENGTH OF PVD COATING

Due to high importance of the fatigue damage of many elements in operation, the study of resistance and degradation mechanism of thin, hard-coated steels has been carried out since the 1980s of last century. Performed studies do not show unambiguous effect of PVD coatings deposition on fatigue resistance, however, some correlations have been found.

Fatigue investigations of systems of hard PVD coating-substrate have shown that kind of fatigue tests and parameters, at which these fatigue tests are performed, coating properties (e.g., thickness, hardness, elastic modulus, and grain size), and also substrate properties have an influence on the obtained fatigue results.

Early studies of an effect of deposition of PVD coatings on fatigue resistance showed an increase of fatigue strength under the tests of $R = -1$, as compared with that of uncoated specimen [12,87–98] and an decrease of fatigue strength under the tests of $R = 0$ or 0.1 [88,93,99], except for the region of low stress amplitude [88,93,100–102]. However, some exceptions from this rule have been observed. The positive effect of PVD of coatings on the fatigue life under the tests of $R = 0.1$ or $R = 0$ was obtained in Baragetti et al. [100], Kim et al. [58], Lecis et al. [103], Sundaram [104], and Zhou et al. [102], while a decrease of fatigue strength of TiN and TiN/Ti multi-layer coatings under the tests of $R = -1$ was obtained in Cassar et al. [105] and Kolkman [106]. It should be noted that in Kolkman [106], the thickness of tested coating was over $40\text{ }\mu\text{m}$, while in Baragetti et al. [100], Ferreira et al. [88], Hotta et al. [89], Kim et al. [58], Shiozawa et al. [93], Su et al. [94–96], Trapezon and Lyashenko [97], Xin et al. [98], and Zhou et al. [102], where the positive effect was obtained, coating thickness was in range of $1\text{--}15\text{ }\mu\text{m}$. This confirms that in addition to the test parameters, the thickness of coating has also a great influence on fatigue resistance.

However, investigations of an effect of thickness of PVD coating on fatigue strength do not give unambiguous result. According to Murakami et al. [107] and Kim et al. [58], the fatigue life increased with increasing film thickness in the range of $2\text{--}6\text{ }\mu\text{m}$. On the other hand, investigations performed by Su et al. [94] and Sundaram [104] showed the decrease of fatigue life of TiN-coated steel with increasing film thickness in the range of $3\text{--}7\text{ }\mu\text{m}$. Investigations shown in Su et al. [94] and Sundaram [104] are in complete opposition to that shown in Murakami et al. [107]. Trapezon and Lyashenko's investigations [97] showed that the influence of coating thickness on fatigue resistance is dependent on deposition parameters and structure of coating (the grain size and number of discontinuities of the deposited coating). An increase of thickness of TiN coatings in the range of $1\text{--}15\text{ }\mu\text{m}$ deposited at a temperature $T = 550\text{ }^{\circ}\text{C}$ and pressure $P = 4.0 \times 10^{-1}\text{ Pa}$ caused an increase of fatigue resistance, but in case of coatings deposited at a lower temperature (500 and $450\text{ }^{\circ}\text{C}$) an increase of coating thickness had an opposite effect, despite the deposition pressure ($3.5 \times 10^{-1}\text{ Pa}$ and $5.0 \times 10^{-1}\text{ Pa}$, respectively) [97]. The relations between a percentage increase of fatigue endurance and coating thickness based on Trapezon and Lyashenko's investigations is shown at Figure 16.10.

**FIGURE 16.10**

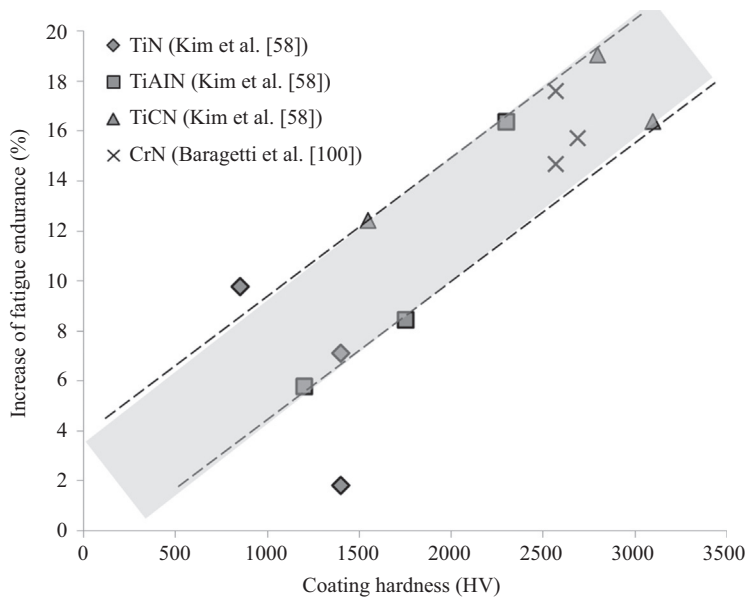
Correlations between an increase of fatigue endurance and coating thickness based on data from Ref. [97].

Kim et al. [58] testing TiN, TiAlN, and TiCN coatings with thickness in the range of 2.5–5 μm have obtained that the best fatigue strength was reached at different coating thicknesses, depending on the coating composition. In case of the TiN and TiAlN coatings, fatigue strength increased with increasing coating thickness, while in case of the TiCN coating, the best fatigue strength was for the coating with the thickness of 3.5 μm. Further increase of coating thickness caused a decrease in fatigue strength. However, all tested PVD coatings in Kim et al. [58], Su et al. [94], and Trapezon and Lyashenko [97] improved fatigue limit in comparison to uncoated substrates.

Coating thickness influences also cracking behavior of PVD coatings. Ganne et al. [108] have observed that in thin PVD coatings (thickness below 3 μm) under tensile and in four-point bending tests, cracks follow the grain boundaries of the substrate. In thicker coatings, cracks were straight, parallel to each other and perpendicular to the tensile axis. This mode I cracking behavior of PVD coatings was also observed under bending and tensile loadings [94,95,109]. The same cracking mode was also observed under thermal cycling test [41].

Experimental investigations to study the influence of coatings properties on fatigue strength have showed that fatigue life of PVD coated steels is related to

- the coating hardness [58,99,102],
- the structure of coating [87,95–97,102],
- adhesion to substrate [99],
- the substrate hardness and residual stress in the substrates [88,89], and
- the ratio between the hardness of the coating and the substrate [100,103,110,111].

**FIGURE 16.11**

Correlation between coating hardness and improvement of fatigue endurance based on Refs. [58,100].

With increasing coating hardness, the fatigue life of coated specimens increased as well (Figure 16.11) [58,87,100,102]. However, according to Kim et al. [58], with increasing coating hardness up to 2800 HV the fatigue strength increased, but after exceeding this hardness a decrease in fatigue endurance was noticed. This shows that there exists a certain coating hardness, above which fatigue strength decreases. This hardness limit is connected with the fact that very hard coatings are often also very stiff and can fracture in a brittle mode, especially if the substrate is much softer than the coating. Experimental observations confirmed that if very hard coating is deposited on soft substrate, coating cannot sustain the substrate elastic strain and cracks in multiple points rapidly arise and grow causing a decrease of fatigue limit [100,105,112]. Thus, the substrate hardness, the mismatch of coating/substrate hardness are more important for fatigue life of coated specimens than the coating hardness alone [99,100,104,105,110]. In the experiments, in which aluminum alloys [100,112] or Ti-6Al-4V titanium alloy [99,105,106] were used as substrates, deposition of hard PVD coatings always caused a drop of fatigue endurance.

Hotta et al. [89] attributed the improvement of number cycles to failure of coated steels in rotating bending tests to an increase of the substrate hardness together with the simultaneous reduction of tensile residual stress in the substrate. The existence of hard, crack-free, and well-bonded coatings deposited on the hard substrates is expected to be effective in preventing the crack initiation and development. Thus,

if the mismatch of coating/substrate hardness and modulus increases, the positive influence of PVD coating deposition decreases leading to a drop of fatigue life below that of uncoated steel.

The improvement of fatigue strength was explained by the process of crack initiation and its development. According to Hotta et al. [89], a hard PVD coating causes suppression of the crack initiation in the substrates by a decrease of the deformation of the substrate surface. Similarly, Kim et al. [58] have observed that under testing at lower stress level, crack initiation is retarded by a hard coating layer on the substrate surface, which acts as a barrier to slip deformation of the plain surface.

Depending on the stress amplitude, different processes in fatigue fracture occurs. When low amplitude stress is applied, fatigue fracture process occurred as a consequence of the propagation of a single crack [87,103,106,113]. This crack might be initiated in a substrate [87,100,103,113] or at a coating/surface interface [106]. The nucleation of the fatigue crack is initiated in a substrate if it is connected with non-metallic inclusions located inside the steel substrate [100,113]. Fracture nucleates at the inclusion due to the stress concentration induced by the metallurgical discontinuity, which propagates both toward the surface of the specimen and the inner portion of the steel, causing the typical fish eye morphology on the fracture surface. However, if substrate is made from steel characterized by high level of cleanliness, than cracks nucleate at the coating [100]. Thus, to improve the fatigue limit by the deposition of PVD coating, a good adhesion is mandatory. A hard well-bonded PVD coating is effective in preventing the crack initiation in the substrates.

In the case of the high amplitude of the stresses, fatigue fracture develops by a co-propagation of multiple cracks initiated at the coating surface, especially at the defects that are spots of the stress concentration. PVD coatings are ceramic materials that possess high strength and low toughness. As a result, PVD coatings are too brittle and stiff to accommodate large elastic strain imposed by substrate deflections under load, so that multi-cracks propagate quickly inside the coating and merge on a fracture plane of specimen leading to a decrease of fatigue life of the substrate [58,88,100,101]. The other place of crack initiation is the coating surface at local micro-defects, such as micro-pits and voids [91,92,94–96,104,105,112]. Each micro-defect acts as a center of stress concentration that favors crack initiation.

Cassar et al. [105] pointed that the main reason of the decrease of fatigue strength is delamination caused by the substantially different elastic moduli of the substrate and coating. Thus, the main reason of lowering the fatigue strength was the mismatch of coating/substrate deformation causing coating delamination and creation of voids at coating-substrate interface, and micro-defects of the coating.

7 DEGRADATION CAUSED BY DYNAMIC LOADING (EROSION)

Erosion of materials (cavitation erosion, liquid impact erosion, and solid particle erosion) is caused by repeated action of dynamic loads on small area of exposed material. This loading may last only few microseconds or even nanoseconds with a high

local-impact pressure of GPa. Material resistance to deformation and fracture under dynamic loading is lower in general than that under quasi-static ones. It has been found that bulk materials, that have good fatigue strength, often also have good erosion resistance. This relationship was found to be also true in the case of PVD coatings [33,44–46,50,56,75,114–118].

According to Gerdes et al. [114], TiN coating deposited on Ti-6Al-4V improved an incubation period approximately three times and a cumulative volume loss decreased ~ 30 times in water droplet erosion. TiN PVD coatings, regardless of the deposition techniques (cathodic arc, electron beam, unbalanced magnetron sputtering), were the most resistant against solid particle erosion [56]. In Mann et al. [119] was shown that TiAlN PVD coating was very effective in preventing degradation against slurry erosion. However, investigations of cavitation erosion resistance showed that in case of very intensive erosion deposition of PVD coating caused mainly an increase of incubation period [120]. It should be added that all coating defects such as micro-droplets and pinholes act as nucleation centers for wear leading to reduced incubation time [45,121]. Many micro-droplets are removed during erosion process leaving small pinholes on the coating surface. All surface defects are the spots of stress concentration that favor spots of cracks initiation. However, a large number of tiny micro-pinholes does not affect a crack initiation and an increase of coating degradation [122].

Similarly to bulk materials, erosion resistance of PVD coatings depends on many factors, such as mechanical properties of coating and substrate (hardness, Young's modulus, tensile strength, and fatigue strength), thermal properties of coating and substrate, microstructure (grain sizes, phases, number and size of defects) and a number of layers in PVD coating, adhesion of a coating to substrate, coating thickness and also on surface roughness. A huge number of factors influencing the resistance of PVD coating to erosion make it impossible to determine the impact of individual factors on the overall behavior of the coating and substrate under the dynamic loading in micro-regions. However, some attempts were made in the literature.

Jianxin et al. [50] have shown that the resistance to solid particle erosion increases with increasing coating hardness and adhesion (critical load). On the other hand, investigations shown in Iwai et al. [117], Krella and Czyżniewski [46], and Krella [44] cannot confirm such relation. As in the case of fatigue, there is a coating hardness, above which erosion resistance decreases [44]. In Krella [44] was performed an attempt to determine an effect of the properties of a coating and a substrate on the resistance of the entire coating-substrate system against cavitation erosion, but only arc-evaporated coatings were investigated.

As it was mentioned earlier, coating ability to resist erosion degradation depends on the substrate hardness. In case of quite soft substrate, for example, austenitic steel, degradation of PVD coatings under dynamic load in cavitation erosion begins from the coating-substrate undulation. This undulation was caused by plastic deformation of the substrate and shear or/and sliding of columnar grains of coating. Undulation of substrate and cyclic strain of coating caused by short lasting loading and elastic behavior of a coating (especially a monolayer coating) may lead to the coating

delamination. The coating delaminated under cyclic dynamic loading easily undergoes shear fracture through the entire coating thickness that ends with coating removal. Another effect of coating undulation is an initiation and increase of tensile stresses in a coating at the bottom and top of undulation—“hills.” These stresses lead to crack initiation and its progress. However, delamination contributes to the stress release at the bottom of hills and undulations. Delaminated coating fractures easily under impact. In Figure 16.12 is shown fracture of TiN coating caused by cavitation. This indicates that coating fractures in delaminated parts are in I and II mode (by tensile and by shearing), and at the tops of “hills,” they are mostly in I mode (by tensile).

In the case of hard substrate, the undulation is slight or sometimes does not occur [54]. However, cyclic loading causes cyclic strain of the coating and substrate and generates complex stress state in the coating. After a number of cycles, stress state in weak points of coating (places with some defects) may exceed the coating strength and may lead to coating fracture. Thus, cracks are caused mostly by shearing. This shows that the erosion behavior of the coating is strongly influenced by the substrates hardness. With increasing substrate hardness, fracture mechanism is changing from I mode to II mode of cracking.

Investigations of an improvement of erosion resistance by applying multi-layer coating instead of monolayer do not give unambiguous results. According to Bromark et al. [75] and Gachon et al. [123], erosion or wear rate decreased if multi-layer coatings were used instead of monolayer ones, while according to Iwai et al. [117] applying of multi-layer coatings may give an opposite result. The decrease of the solid particle erosion resistance of multi-layer coatings was explained by a very small scale of the wear fragments and a very low thickness of soft layers that played the role of the defects that facilitate a removal of coating fragments.

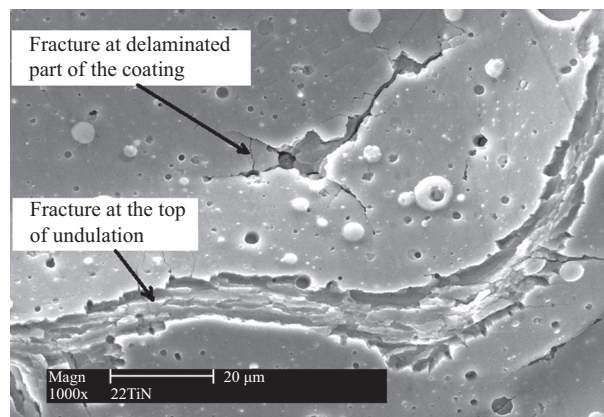


FIGURE 16.12

Fracture of a TiN coating deposited on austenitic steel caused by cavitation.

Krell and Czyżniewski [54], permission from Elsevier.

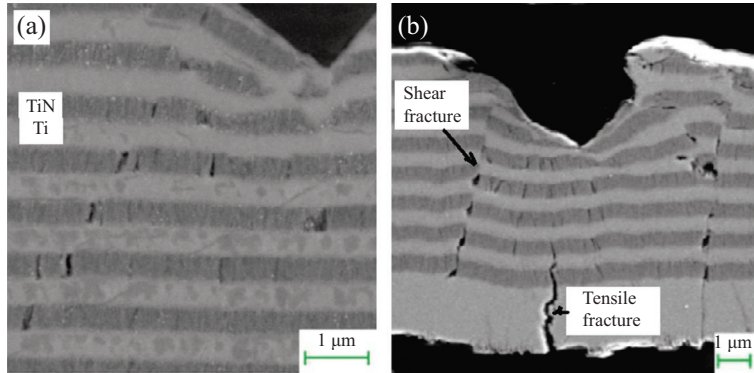


FIGURE 16.13

Shear fracture in Ti/TiN multi-layer coatings caused by solid particle erosion.

Maurer and Schulz [79], permission from Elsevier.

In case of degradation of Ti/TiN multi-layer coating caused by solid particle erosion, shear fractures initially occurred in harder TiN ceramic layers (Figure 16.13a), and then they propagated through soft Ti layers (Figure 16.13b) [79]. Fracture of PVD coating is mainly determined by its columnar structure, the coating is prone to fail along those columns. Thus, the degradation mechanisms were similar to these developed during quasi-static indentation (see Section 4). Moreover, Figure 16.13 shows that the substrate hardness influences stress distribution and fracture mechanics similarly to cavitation erosion. A substrate and coating undergo plastic deformation and indentation under impact of solid particle and tensile stress is introduced to the coating leading to crack initiation. Crack starts at the substrate-Ti layer interface and grows in the direction to the top of coating.

8 CONCLUSION

There was presented an influence of deposition parameters on PVD coating properties. An influence of bias voltage, which has an effect on most structural and mechanical properties of coating, was analyzed in detail. It was shown that substrate properties, such as substrate hardness, elastic modulus, and thermal properties also influence the PVD coating properties. In the case of multi-layer coating, its properties depend on the combination of properties of all layers and the number and thickness of each layer.

Deformation of monolayer coating occurs via shearing and sliding at intercolumnar boundary under the action of shear stresses. In case of multi-layer coating, soft layers act as buffer layers, and cracks, in general, initiate at interfaces and propagate through a hard ceramic layers.

Although a lot of investigations were performed in the literature, in which same relation between deposition parameters and coating properties, and also between coating properties and erosion resistance were obtained, and in most cases it was proved that the deposition of PVD coatings improves fatigue strength and erosion resistance, any specific PVD coating-substrate systems exposed to a specific kind of erosion or fatigue requires further detailed experimental studies.

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Application of pyrolysis-gas chromatography/mass spectrometry (Py-GC/MS) and scanning electron microscopy (SEM) in failure analysis for the identification of organic compounds in chemical, rubber, and automotive industry

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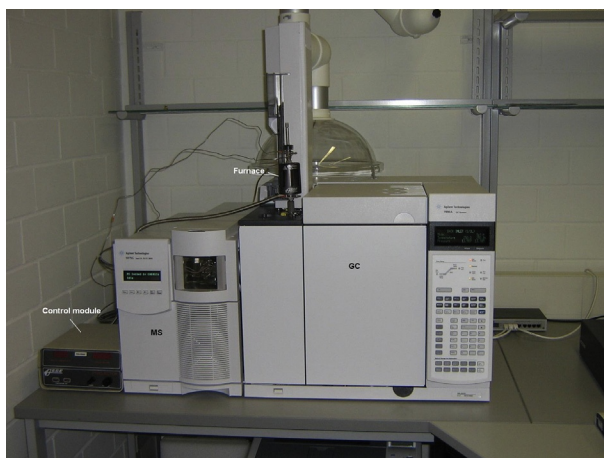
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1 INTRODUCTION

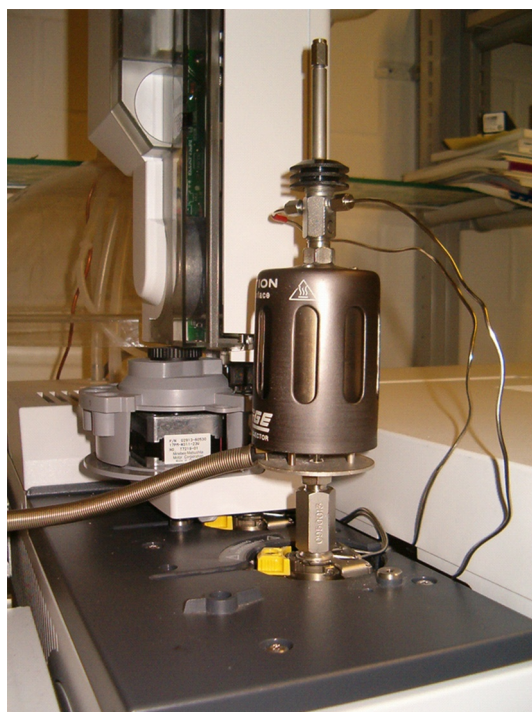
Failure of the structure of materials or components often results in accidents and plant shutdowns, resulting in hefty compensations. Failure analysis is the process of collecting and analyzing data to determine the cause of a failure and to take action to prevent it from reoccurring. It is an important discipline in many branches of manufacturing industry, such as the automotive industry and chemical or rubber industry.

Various analytical techniques, like microscopy imaging, scanning electron microscopy (SEM), energy dispersive X-ray analysis, UV/Vis spectrometry, Fourier transform infrared spectroscopy (FTIR), nuclear magnetic resonance, time-of-flight secondary ion mass spectrometry, and others, are used for clearing of failure of raw material, manufacturing, function, design or storage errors of various plastic or metal components from the automotive or chemical and rubber industry. SEM is a technique that utilizes an electron beam to produce magnified images of samples [1]. The sample surface is scanned with a beam of energetic electrons and a number of processes occur that can be used to generate an image of the sample surface. The SEM allows the exact localization of damage of components from engineering and automotive industry.

The analytical pyrolysis-gas chromatography/mass spectrometry (Py-GC/MS) technique has extended the range of possible tools for characterization of synthetic polymeric materials. Under controlled conditions at elevated temperature (500-1400 °C) in the presence of an inert gas, reproducible decomposition products characteristic for the original polymer/copolymer sample are formed [2-5]. The pyrolysis unit (pyrolyzer) is directly connected to the injector port of a gas chromatograph (GC) (Figures 17.1 and 17.2). Flow of an inert carrier gas, such as helium, flushes the pyrolyzates into the fused silica capillary column, which is installed in the thermostat of the GC (Figure 17.1). The capillary column separates the complex mixture of the organic substances. The chromatographically

**FIGURE 17.1**

Pyrolysis-GC/MS system used in this work equipped with a furnace pyrolyzer *Pyrojector II*TM (SGE Analytical Science), control module (SGE Analytical Science), a 7890A gas chromatograph (GC), and a series 5975C quadrupole mass spectrometer (MS) (both from Agilent Technologies).

**FIGURE 17.2**

View of the furnace pyrolyzer *Pyrojector II*TM (SGE Analytical Science) connected to the injection port of the gas chromatograph (GC).

separated substances are then detected with mass spectrometer (MS) (Figure 17.1), and subsequently identified by interpretation of the obtained mass spectra or by using mass spectral libraries (e.g., NIST/EPA/NIH, Wiley, MPW, or Norman Mass Bank). The detection technique of the separated compounds is typically MS, but other gas chromatographic detectors have been also used depending on the intentions of the analysis. The analytical Py-GC/MS allows the direct analysis of very small solid or liquid polymer/copolymer sample amounts (5–200 μg), without the need of time-consuming sample preparation (sample pretreatment). The identification of complex mixtures or blends as well as identification of samples with so-called difficult matrices is also possible in many cases. Due to these small sample amounts, the investigation of heterogeneous polymers with a coarse phase or a gradient composition structure (phase separation, poor mixing, etc.) is sophisticated and may lead to great variations in the measuring results. In this case, a multiple determination of different positions of the investigated part is essential to achieve a significant image of its composition.

The applications of the analytical Py-GC/MS range from research and development of new materials, quality control, characterization and competitor product evaluation, medicine, biology and biotechnology, geology, airspace, environmental analysis to forensic purposes or conservation and restoration of cultural heritage. These applications cover analysis and identification of polymers/copolymers and additives in components of automobiles, tires, packaging materials, textile fibers, coatings, adhesives, half-finished products for electronics, paints or varnishes, lacquers, leather, paper or wood products, food, pharmaceuticals, surfactants, and fragrances [2–5].

For more than 10 years, our laboratory has been involved in projects from the area of failure analysis in the automotive, chemical, or rubber industry by using the analytical Py-GC/MS. The practical application of these analytical techniques ranges from case studies of automotive components failure analysis of failed hydraulic cylinders, membranes, as well as the packaging of hydraulic cylinders, sealing rings, tire materials and additives, auto paints, auto wrapping foils, auto catalysts, mineral oils, and brake fluids. Further applications include identification of plastic particles from industrial filter fins, residues of fittings of refrigerant compressors, residues from a valve block of a medical respirator, identification of dental filling materials, detection of counterfeits of plastic and rubber products, modules from building industry, fouling from a roller bearing, adhesives, sealing compounds, cesspit residues from an extruder or polyethylene regranulate from mechanical recycling process, or polymer residues in recycled aluminum. The obtained analytical results are then used for troubleshooting and remedial action of the technological process.

Here, we report on application examples of the analytical Py-GC/MS and the SEM in failure analysis of chemical materials such as motor vehicle brake fluids, varnishes, and rubber materials of car tires. Furthermore, failure cases demanding identification of polymers/copolymers in paint on the surface of a rubber membrane from the pressure vessel of an automotive integral accumulator and identification of

the composition of sludge on the surface of the compressor wall from car air conditioner are demonstrated. In addition, the identification of scaling of a valve block unit from a failed medical oxygen resuscitation apparatus is presented.

2 EXPERIMENTAL

2.1 SAMPLES

Passenger car tires, car brake fluids, varnishes, rubber membrane from the pressure vessel of an automotive integral accumulator, a part from car air conditioner, and a valve block unit from the medical oxygen resuscitation device were used in the investigation.

Poly(vinyl alcohol) (PVA) obtained from Celanese Chemicals Europe GmbH (Frankfurt/Main, Germany) and polyvinylpyrrolidone (PVP) (*Luvitec K 30*) obtained from BASF AG (Ludwigshafen, Germany) were used as standards for qualitative identification of samples.

2.2 INSTRUMENTATION AND ANALYTICAL CONDITIONS

Approximately 100–200 µg of a solid sample were cut out with scalpel and inserted without further preparation into the bore of the pyrolysis solids-injector and then placed with a plunger on the quartz wool of the quartz tube of the furnace pyrolyzer *Pyrojector II*TM (SGE Analytical Science, Melbourne, Australia) (Figure 17.2). The pyrolyzer was operated at constant temperature of 550 or 700 °C. The pressure of helium carrier gas at the inlet to the furnace was 95 kPa.

Py-GC/MS measurements were made by using two apparatuses. In the first apparatus (1), the pyrolyzer was connected to a *Trace 2000* GC (ThermoQuest/CE Instruments, Milan, Italy) with a quadrupole MS *Voyager* (ThermoQuest/Finnigan, MassLab Group, Manchester, UK) operated in electron impact ionization (EI) mode. The fused silica GC capillary column (*Elite-5ms* 60 m long, 0.25 mm I.D., 0.25 µm film thicknesses; PerkinElmer Instruments, Shelton, CT, USA) was used. The gas chromatographic conditions were as follows: programmed temperature of the capillary column from 60 °C (7 min hold) at 5 °C min⁻¹ to 100 °C and then 10 °C min⁻¹ to 280 °C (25 min hold at 280 °C). The temperature of the split/splitless injector was 250 °C and the split flow was 10 cm³ min⁻¹. Helium, grade 5.0 (Westfalen AG, Münster, Germany) was used as a carrier gas at constant pressure of 70 kPa during the whole analysis. The transfer line temperature was 280 °C. The MS EI ion source temperature was kept at 250 °C. The ionization occurred with a kinetic energy of the impacting electrons of 70 eV. The current emission of the rhenium filament was 150 µA. The MS detector voltage was 350 V. Mass spectra and reconstructed chromatograms (total ion current (TIC)) were obtained by automatic scanning in the mass range *m/z* 35–450 u. Py-GC/MS data was

processed with *Xcalibur* software (ThermoQuest) and the *NIST 05* mass spectral library. In the second apparatus (2), the pyrolyzer was connected to a 7890A GC with a series 5975C quadrupole MS (Agilent Technologies Inc., Santa Clara, CA, USA) operated in EI mode (Figure 17.1). The fused silica GC capillary column (60 m long, 0.25 mm I.D. with *DB-5ms* stationary phase, film thickness 0.25 μm ; J&W Scientific, Folsom, CA, USA) was used. Helium, grade 5.0 (Westfalen AG) was used as a carrier gas.

The gas chromatographic conditions were as follows:

1. Programmed temperature of the capillary column from 60 °C (1 min hold) at 7 °C min^{-1} to 280 °C (hold to the end of analysis) and programmed pressure of helium from 122.2 kPa (1 min hold) at 7 kPa min^{-1} to 212.9 kPa (hold to the end of analysis),
2. Programmed temperature of the capillary column from 75 °C (1 min hold) at 7 °C min^{-1} to 280 °C (hold to the end of analysis) and programmed pressure of helium from 122.2 kPa (1 min hold) at 7 kPa min^{-1} to 212.9 kPa (hold to the end of analysis).

The temperature of the split/splitless injector was 250 °C and the split ratio was 20:1. The transfer line temperature was 280 °C. The MS EI ion source temperature was kept at 230 °C.

The ionization occurred with a kinetic energy of the impacting electrons of 70 eV. The quadrupole temperature was 150 °C. Mass spectra and reconstructed chromatograms (TIC) were obtained by automatic scanning in the mass range m/z 35–750 u. Py-GC/MS data were processed with the *ChemStation* software (Agilent Technologies) and the *NIST 05* mass spectral library.

The scanning electron microscopic studies were carried out using the JSM-6510 series JEOL (Germany) GmbH (Eching/München) apparatus.

2.3 DERIVATIZATION AND DIRECT GC/MS ANALYSIS

Micro-reaction vessels of 3 cm^3 with solid cap and PTFE liner (Supelco, Bellefonte, PA, USA) along with the block heater SBH 200DC (Stuart, Bibby Scientific Ltd., Stone, Staffordshire, UK) were used for derivatization. Approximately 100 μL of the investigated brake fluid sample was placed in the micro-reaction vessel and 300 μL of the silylation agent *N*-methyl-*N*-(trimethylsilyl)-trifluoroacetamide (MSTFA) was added. The sealed vessel was placed in the block heater at 70 °C for 30 min. The resulting product was dissolved 1:1 (v/v) in methanol and 1 μL of the solution was injected into the GC/MS apparatus for analysis. Programmed temperature of the capillary column *DB-5ms* (30 m long, 0.25 mm I.D., 0.25 μm film thickness) (J&W) from 60 °C (1 min hold) at 6 °C min^{-1} to 280 °C (hold to the end of analysis) was used for GC separation of compounds. Helium, grade 5.0 (Westfalen AG) was used as a carrier gas at constant flow rate of 1 mL min^{-1} . The temperature of the split/splitless injector was 250 °C and the split ratio was 25:1. The transfer line temperature was 280 °C. The MS EI ion source temperature was

kept at 230 °C. The ionization occurred with a kinetic energy of the impacting electrons of 70 eV. The quadrupole temperature was 150 °C. Mass spectra and reconstructed chromatograms (TIC) were obtained by automatic scanning in the mass range m/z 35-750 u. The solvent delay was 9 min.

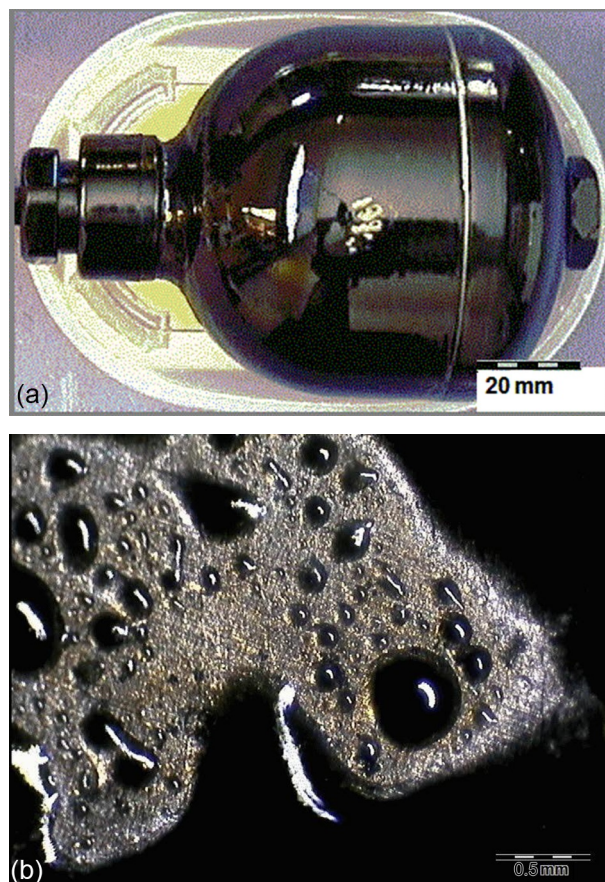
3 RESULTS

3.1 APPLICATION OF Py-GC/MS AND SEM IN FAILURE ANALYSIS IN THE AUTOMOTIVE INDUSTRY

For the failure analysis in motor vehicles, there is often lack of information about the component itself, such as chemical composition, temperature resistance, possible contaminants, or mechanical properties. The damage range is usually limited and not always homogeneous. The SEM allows the exact localization of damage of components from engineering and automotive industry. There are often only small amounts of samples available to clarify the damage, which may be important for recognizing the cause of damage [2,3]. The increasing use of polymeric materials in the automotive industry requires sensitive and reliable methods for its analysis. Traditional analytical techniques used for characterization of polymers/copolymers, such as thermal analysis and FTIR, have limitations or are not sufficiently sensitive to demonstrate the change of the structure and the resulting dysfunction of used materials. A lot of information about dysfunction of automobile parts can be obtained from the fouling material on the surface of the failed parts. In such cases, the sampling can be made by rubbing the affected surface with the quartz glass wool followed by the Py-GC/MS of the enriched wool. The other problem could be the difficulty in the interpretation of pyrograms (Py-GC/MS chromatograms) of complex mixtures or blends because of the large number of decomposition products. In such cases, an extended analysis is required.

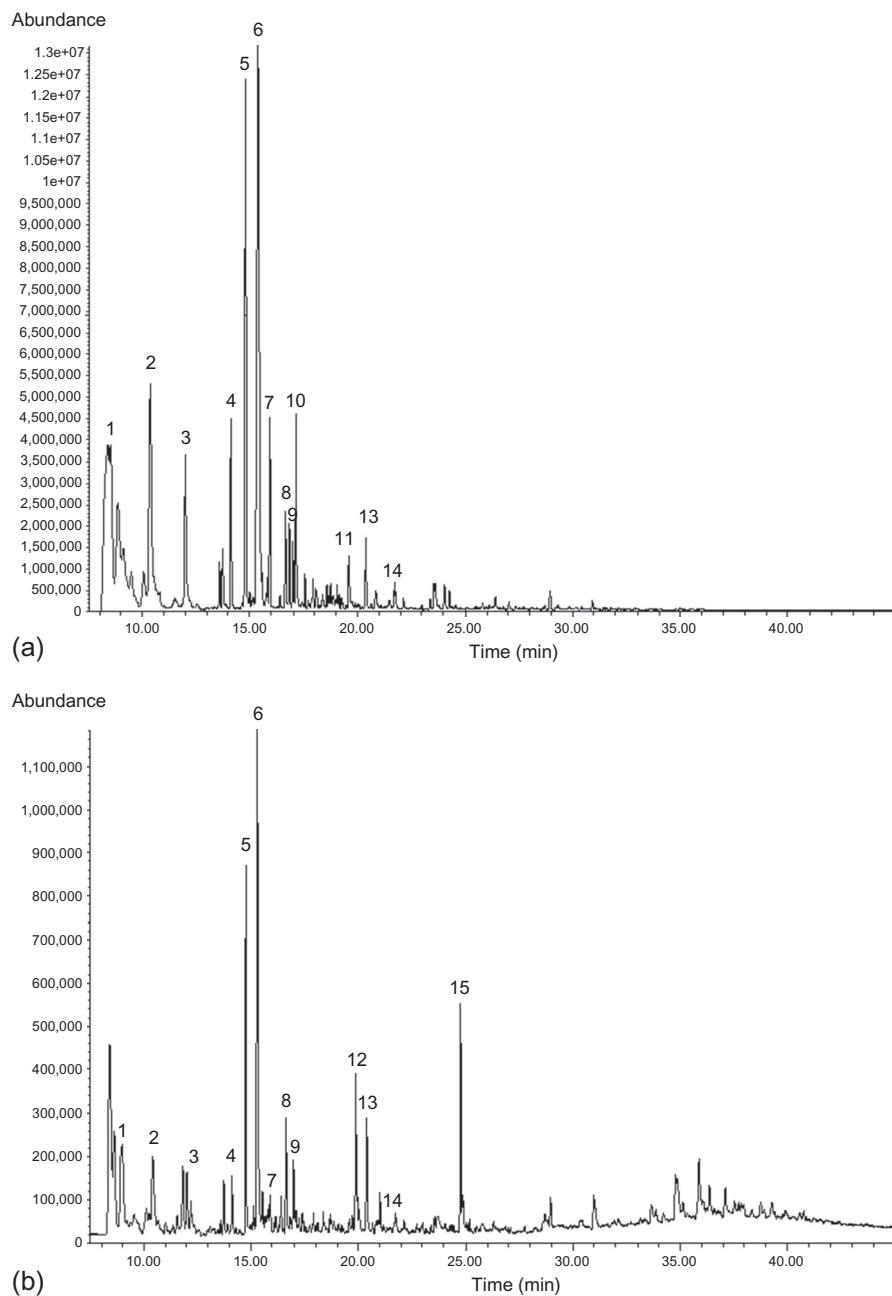
3.1.1 Case 1: Defect work in paint on the surface of a rubber membrane from the pressure vessel of an automotive integral accumulator

The goal of the first study was to investigate the cause of defect work in paint on the surface of a rubber membrane from a pressure vessel (automotive component). Figure 17.3a and b shows the state of the investigated pressure vessel and a magnified view of the defect location on the rubber membrane surface, respectively. The samples of the lacquer coating from the good and from the defective surface of the rubber membrane were pyrolyzed at 550 and 700 °C, respectively followed by GC/MS analysis to identify their composition. Figures 17.4 and 17.5 show the obtained pyrograms of the samples. Based on the decomposition products summarized in Table 17.1, the samples were identified as a mixture of styrene-butadiene rubber (SBR), phenol resin, poly(ethylene terephthalate) (PET), and additives. The main feature of the pyrolysis of SBR is the formation of butene/butadiene (RT = 8.60 min), benzene (RT = 10.37 min), toluene (RT = 12.02 min), styrene (RT = 14.14 min), and naphthalene (RT = 19.60 min)

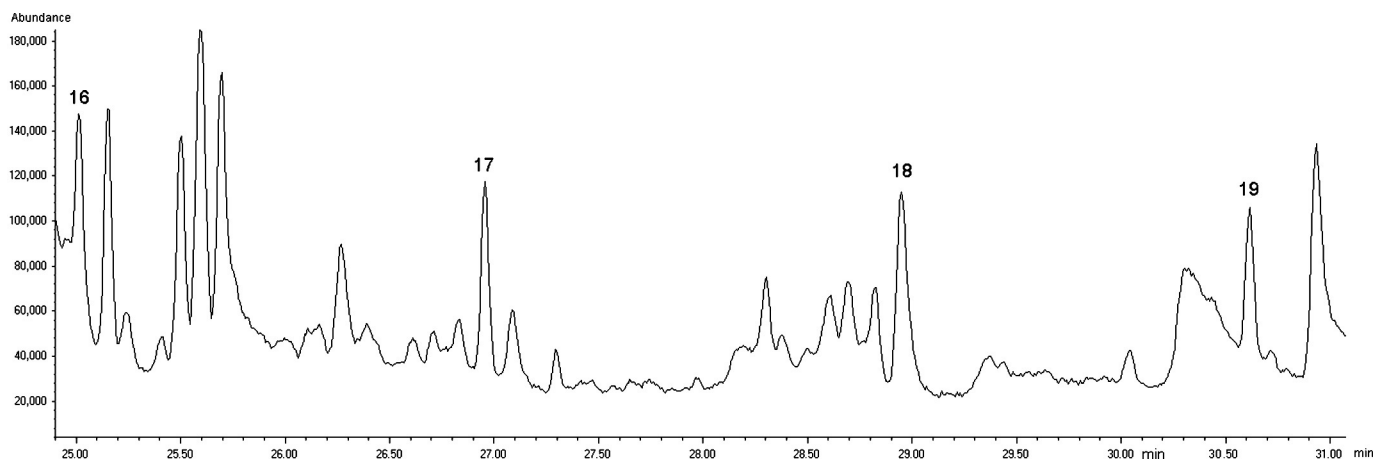
**FIGURE 17.3**

(a) Imaging of the investigated rubber membrane from the automotive pressure vessel.
 (b) Imaging of the defective surface of the rubber membrane from the automotive pressure vessel.

(see Table 17.1). The presence of phenol ($RT = 15.40$ min) and cresol isomers ($RT = 16.64$ and 16.99 min) in the pyrolyzate is characteristic for the thermal degradation of phenol resin. The peaks of benzofuran ($RT = 15.87$ min) and acetophenone ($RT = 17.13$ min) were assigned to PET. Other substances that appear in pyrograms in Figure 17.4 have been identified by EI MS as additives or their thermal degradation products. The peak of benzothiazole ($RT = 20.40$ min) corresponds to the product of the thermal decomposition of benzothiazole-2-thiol (2-mercaptobenzothiazole, 2-MBT). In the rubber industry, benzothiazole-2-thiol is used as vulcanization accelerator and as antioxidant. One of the main uses of 2-phenoxyethyl acrylate, identified at $RT = 24.74$ min, is in coatings on glass, metal, paper, plastic, wood, and PVC floor coatings. Phenoxyethanol identified at $RT = 19.88$ min is used upon other applications as solvent and plasticizer for

**FIGURE 17.4**

Comparison of pyrolysis-GC/MS chromatograms at 700 °C of (a) good surface and (b) defective surface of the rubber membrane from the automotive pressure vessel. Pyrolysis-GC/MS apparatus 2, chromatographic conditions 1. For peak identification, see [Table 17.1](#).

**FIGURE 17.5**

Pyrolysis-GC/MS chromatogram at 550 °C (range 24.5–31.5 min) of the defective surface of the rubber membrane from the automotive pressure vessel. Pyrolysis-GC/MS apparatus 2, chromatographic conditions 1. For peak identification, see [Table 17.1](#).

Table 17.1 Pyrolysis Products of the Rubber Membrane with Coating at 550 and 700 °C

Peak	Retention Time (RT) (min)	Pyrolysis Product	Pyrolyzed Material
1	8.60	1-Butene/1,3-butadiene	SBR
2	10.37	Benzene	SBR, phenol resin
3	12.02	Toluene	SBR, phenol resin
4	14.14	Styrene	SBR
5	14.83	Ethenyloxybenzene	Additive
6	15.40	Phenol	Phenol resin
7	15.87	Benzofuran	PET
8	16.64	<i>o</i> -Cresol	Phenol resin
9	16.99	<i>p</i> -Cresol	Phenol resin
10	17.13	Acetophenone	PET
11	19.60	Naphthalene	SBR
12	19.88	Phenoxyethanol (Dowanol EP)	Additive
13	20.40	Benzothiazole	Additive
14	21.73	2-Methylbenzothiazole	Additive
15	24.74	2-Phenoxyethyl acrylate	Additive
16	25.03	1-Pentadecene	Mineral oil
17	26.96	1-Hexadecene	Mineral oil
18	28.95	1-Heptadecene	Mineral oil
19	30.62	1-Octadecene	Mineral oil
20	32.0-45.0	Heavy <i>n</i> -alkanes	Mineral oil

resin coatings. It was assumed that ethenyloxybenzene (RT = 14.83 min) was generated from the additive. Additionally, by pyrolyzing the lacquer coating from the defective surface of the rubber membrane, a mixture of straight-chain aliphatic hydrocarbons corresponding to heavy *n*-alkanes and α -alkenes (Figures 17.4 and 17.5; Table 17.1) was detected. The detected compounds suggest the presence of a mineral oil (heavy *n*-alkanes) on the defective surface of the rubber membrane. The identified α -alkenes (Figure 17.5, peaks 16-19) are known as pyrolysis products of mineral oil [2]. Based on the obtained results, it was found that the cause of the defect in paint work on the surface of a rubber membrane from the automotive pressure vessel was the mineral oil. The obtained analytical results were then used for troubleshooting with the customer.

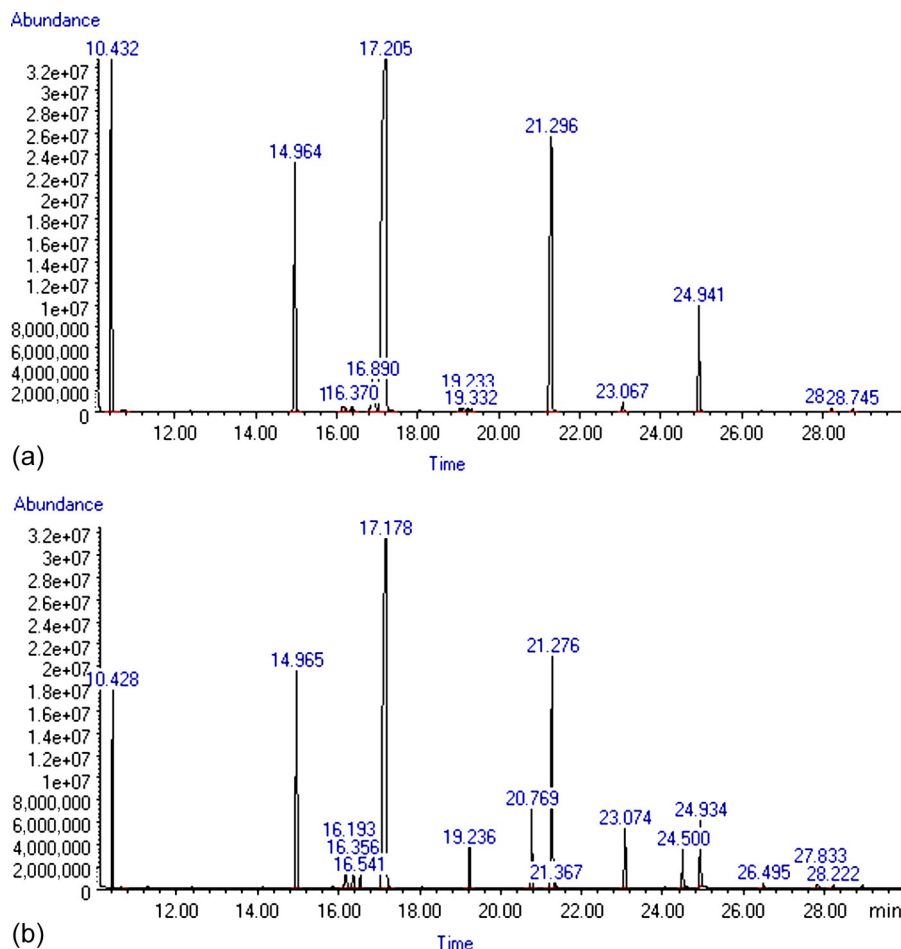
3.1.2 Case 2: Identification of the composition of claimed motor vehicle brake fluid

Motor vehicle brake fluids are used to transfer energy to the actuating mechanism in the hydraulic drive of the brake assembly of a car and to remove heat from the brake discs [6]. The main requirements for brake fluids are high operation temperatures,

good low-temperature and viscosity-temperature properties, physical and chemical stability, protection of metals from corrosion, inactivity with respect to mechanical rubber articles, and lubricating effect [6]. Synthetic brake fluid bases on mixtures of diethylene glycol and di-, tri-, along with tetraethylene glycol monoalkyl ethers. They contain inhibitors to prevent the corrosion of metallic brake components and to reduce oxidation at increased temperatures. To increase the boiling point, tris(methyl glycol) borates are introduced in brake fluids [6].

In this case study, in order to find the difference between the chemical composition of the good functioning and the claimed motor vehicle brake fluids, both samples were silylated and analyzed by direct GC/MS without previous pyrolysis. Silylation is the introduction of a silyl group into a chemical molecule, usually in substitution for active hydrogen in the hydroxyl group (—OH) in alcohols, phenols, carboxylic acids, oximes, sulpho-acids, boric acids, phosphorous acids, in the —NH group in amines, amides, imines, and in the SH group in thiols and thiolcarboxylic acids by a silyl group. Replacement of active hydrogen by a silyl group reduces the polarity of the compound and reduces hydrogen bonding [7,8]. The silylated derivatives are more volatile and thermally stable. The detection of compounds is enhanced. The trimethylsilyl (TMS) group, $\text{—Si(CH}_3\text{)}_3$, is the most popular and versatile silyl group for GC and GC/MS analysis. Nonpolar polymethylphenylsiloxane GC stationary phases, such as the *DB-5ms* phase used in this work, combine inertness and stability with excellent separating characteristics for these derivatives. Figure 17.6 shows the obtained TIC of both investigated brake fluids after trimethylsilylation reaction with *N*-methyl-*N*-(trimethylsilyl)-trifluoroacetamide at 70 °C. The silylated compounds of both brake fluids identified by interpretation of the obtained mass spectra and by using the *NIST 05* mass spectral library are summarized in Table 17.2. The mass spectrum of the trimethylsilylated derivative of triethylene glycol monomethyl ether (main ingredient of the investigated brake fluids) is shown in Figure 17.7. The weak signal at m/z 237 in Figure 17.7 shows the quasi-molecular ion $[\text{M}+\text{H}]^+$ of trimethylsilylated triethylene glycol monomethyl ether (see the structural formula in Figure 17.7). The fragment at m/z 221 corresponds to the ion $[\text{M}-\text{CH}_3]^+$. The fragments at m/z 31, 45, 59, 75, 89, 103, 119, 133, and 147 correspond to the ions $[\text{CH}_3\text{O}]^+$, $[\text{CH}_3\text{OCH}_2]^+$, $[\text{CH}_3\text{OCH}_2\text{CH}_2]^+$, $[\text{CH}_3\text{OCH}_2\text{CH}_2\text{O}]^+$, $[\text{CH}_3\text{OCH}_2\text{CH}_2\text{OCH}_2]^+$, $[\text{CH}_3\text{OCH}_2\text{CH}_2\text{OCH}_2\text{CH}_2]^+$, $[\text{CH}_3\text{OCH}_2\text{CH}_2\text{OCH}_2\text{CH}_2\text{O}]^+$, $[\text{CH}_3\text{OCH}_2\text{CH}_2\text{OCH}_2\text{CH}_2\text{OCH}_2]^+$, and $[\text{CH}_3\text{OCH}_2\text{CH}_2\text{OCH}_2\text{CH}_2\text{OCH}_2\text{CH}_2]^+$, respectively. They occur through a cleavage mechanism of the alkoxy chain in the structure of the identified trimethylsilylated triethylene glycol monomethyl ether molecule. The peaks in the mass spectrum in Figure 17.7 at m/z 73, 89, 103, and 117 are formed from the $[\text{Si(CH}_3\text{)}_3]^+$, $[\text{OSi(CH}_3\text{)}_3]^+$, $[\text{CH}_2\text{OSi(CH}_3\text{)}_3]^+$, and $[\text{CH}_2\text{CH}_2\text{OSi(CH}_3\text{)}_3]^+$ ions, respectively.

As can be seen from Table 17.2, the chemical composition of the claimed brake fluid differs from the composition of the good functioning brake fluid. Although the content of the main component triethylene glycol monomethyl ether in both samples is approximately equal to 46.5%, the claimed brake fluid contains additional

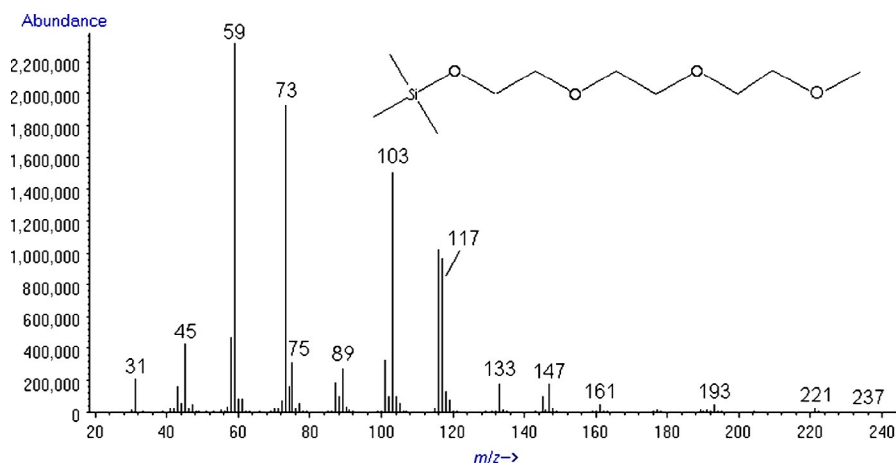
**FIGURE 17.6**

GC/MS chromatograms (TIC) of (a) good functioning motor vehicle brake fluid and (b) claimed motor vehicle brake fluid after derivatization reaction. GC/MS apparatus 2 and analytical conditions as in p. 2.3. For peak identification, see Table 17.2.

compounds, such as tri-, tetra-, and pentaethylene glycol monobutyl ethers. The content of the detected tri-, tetra-, and pentaethylene glycol in the claimed brake fluid sample is greater, while the content of boric acid is by about 7.5% lower. On the other hand, the content of the corrosion inhibitor bis(2-hydroxyethyl)methylamine in the claimed brake fluid is about three times greater as in the good functioning brake fluid. The obtained analytical results were then used for troubleshooting with the customer.

Table 17.2 Identification and Content of the Investigated Motor Vehicle Brake Fluids

Retention Time (RT) (min)	Identified Compound (Trimethylsilyl Derivate)	Content, % (w/w)	
		Good Functioning Brake Fluid	Claimed Brake Fluid
10.43	Boric acid	17.41	9.87
14.96	Diethylene glycol	10.64	11.26
16.36-16.37	Bis(2-hydroxyethyl)methylamine	0.40	1.21
17.18-17.20	Triethylene glycol monomethyl ether	46.37	46.58
19.23-19.24	Triethylene glycol	0.61	1.81
20.77	Triethylene glycol monobutyl ether	-	3.55
21.28-21.30	Tetraethylene glycol monomethyl ether	15.90	14.09
23.07	Tetraethylene glycol	0.28	2.45
24.50	Tetraethylene glycol monobutyl ether	-	1.62
24.93-24.94	Pentaethylene glycol monomethyl ether	3.78	2.94
26.49	Pentaethylene glycol	-	0.21
27.83	Pentaethylene glycol monobutyl ether	-	0.47
	Others	4.61	3.94

**FIGURE 17.7**

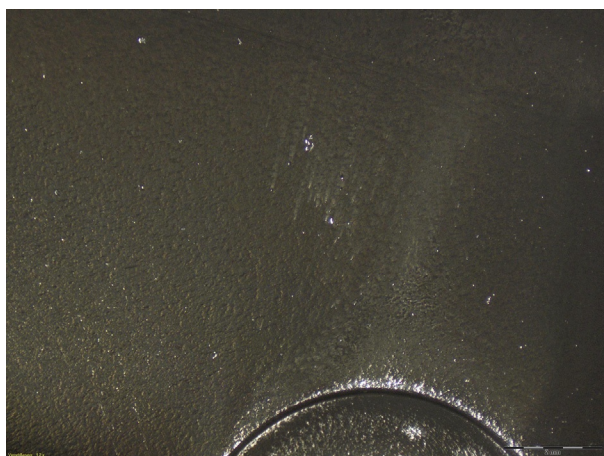
Chemical structure and mass spectrum of trimethylsilyl derivate of triethylene glycol monomethyl ether (main ingredient) identified in the examined motor vehicle brake fluids. GC/MS apparatus 2 and analytical conditions as in p. 2.3.

3.1.3 Case 3: Identification of the composition of sludge on the surface of the compressor wall from car air conditioner

In the following case, sludge on the surface of the compressor wall from a car air conditioner was identified. Figure 17.8a and b shows the fragment of the compressor wall from the car air conditioner and a magnified view of the sludge allocation on the surface of the investigated component, respectively. Figure 17.9 shows the pyrogram of the examined sludge pyrolyzed at 700 °C. The pyrolysis products identified by using mass spectral library *NIST 05* are summarized in Table 17.3. Based on the identified decomposition products, the sludge was identified as a mixture of SBR,



(a)



(b)

FIGURE 17.8

(a) Imaging of the compressor wall from the investigated car air conditioner. (b) Imaging of the sludge on the surface of the compressor wall from the investigated car air conditioner.

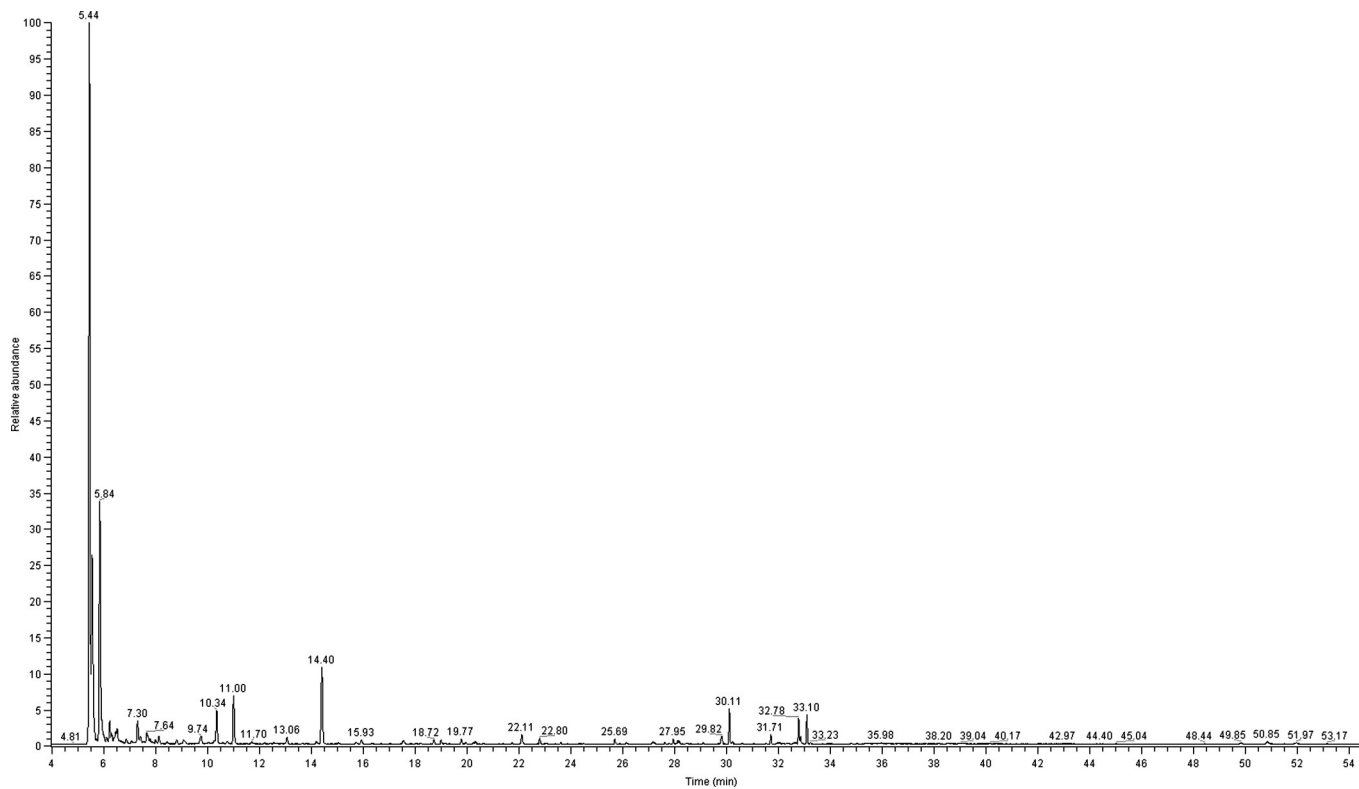


FIGURE 17.9

Pyrolysis-GC/MS chromatogram (TIC) at 700 °C of sludge on the surface of the compressor wall from the investigated car air conditioner. Pyrolysis-GC/MS apparatus 1 and analytical conditions for apparatus 1. For peak identification, see [Table 17.3](#).

Table 17.3 Pyrolysis Products of Sludge on the Surface of the Compressor Wall from Car Air Conditioner at 700 °C

Retention Time (RT) (min)	Pyrolysis Product	Pyrolyzed Material
5.44	Propylene	Polypropylene glycol, SBR
5.54	Acetaldehyde	Polypropylene glycol
5.61	1-Butene	SBR
5.84	Acetone	Polypropylene glycol
6.22	1-Hexene	Mineral oil
7.30	Acetic acid anhydride	Polypropylene glycol
7.64	Methoxyacetone	Polypropylene glycol
8.12	Propylene glycol	Polypropylene glycol
9.74	Toluene	SBR
10.34	Diisopropyl ether	Polypropylene glycol
11.00	Isopropoxy acetone	Polypropylene glycol
13.06	Ethylbenzene	SBR
14.19	1-Decene	Mineral oil
14.40	Styrene	SBR
15.93	Cumene	SBR
18.72	α -Methylstyrene	SBR
19.00	1-Undecene	Mineral oil
22.11	Cresol	Additive
22.80	1-Dodecene	Mineral oil
25.69	1-Tridecene	Mineral oil
27.95	1-Tetradecene	Mineral oil
29.82	1-Pentadecene	Mineral oil
30.11	Undecanal	Additive
31.71	2,6-bis-(1,1-dimethylethyl)-4-methylphenol (BHT)	Additive
32.78	1-Hexadecene	Mineral oil
33.10	Dodecanal (lauric aldehyde)	Additive
49.85	Tri- <i>o</i> -cresyl phosphate	Additive
50.85	Tri- <i>p</i> -cresyl phosphate	Additive
51.97	Tri- <i>m</i> -cresyl phosphate	Additive

polypropylene glycol, mineral oil (paraffin oil), and additives, such as 2,6-bis-(1,1-dimethylethyl)-4-methylphenol (BHT, antioxidant), undecanal, lauric aldehyde (antioxidants), and tricresyl phosphate isomers (flame retardant agents). They are also used as fire-resistant in hydraulic fluids and lubricants [9]. The mass spectrum of the identified tri-*p*-cresyl phosphate is shown in Figure 17.10. The obtained analytical results were then used for troubleshooting with the customer.

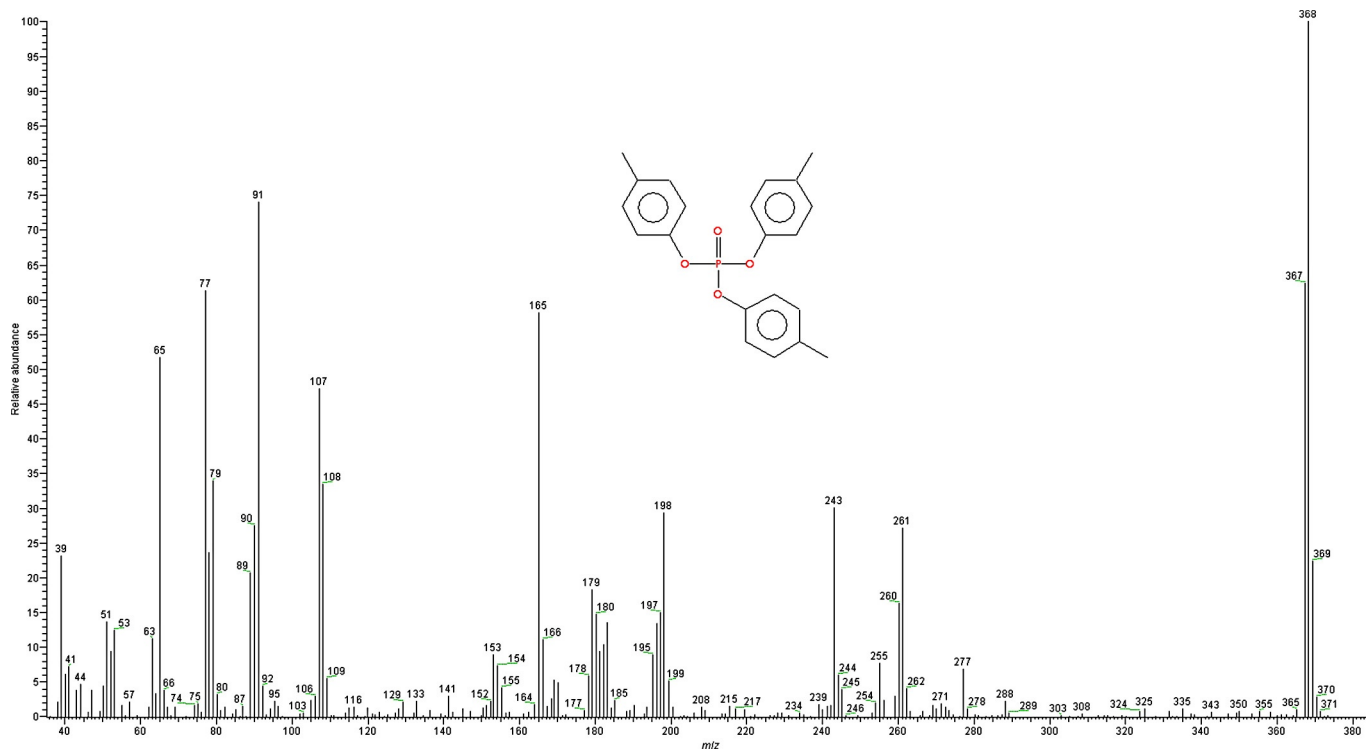


FIGURE 17.10

Chemical structure and mass spectrum of the flame retardant agent tri-*p*-cresyl phosphate identified in sludge on the surface of the compressor wall from the investigated car air conditioner. Apparatus 1 and analytical conditions for apparatus 1.

3.2 APPLICATION OF Py-GC/MS AND SEM IN FAILURE ANALYSIS IN THE RUBBER AND AUTOMOTIVE INDUSTRY

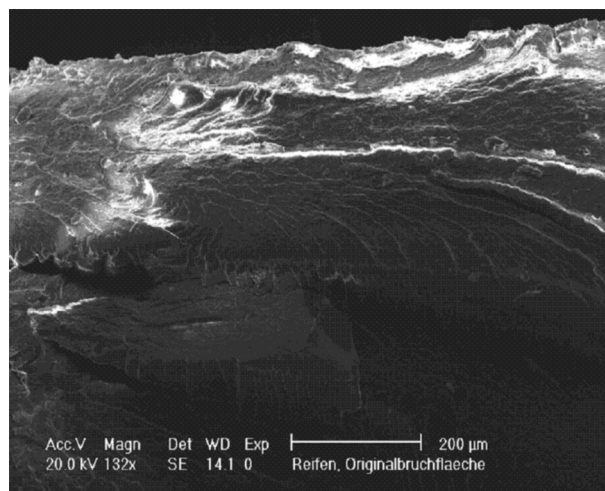
Commonly used rubbers in industry are natural rubber (NR, polyisoprene), synthetic polyisoprene (IR), polybutadiene (BR), styrene-butadiene copolymers (SBR), and nitrile rubber (NBR). Nitrile rubber [poly(acrylonitrile-*co*-butadiene)] is a copolymer containing 15-50% acrylonitrile, manufactured by emulsion polymerization of acrylonitrile and 1,3-butadiene. It was invented at roughly the same time as SBR (near the end of the 1920s) as a substitute for NR [4].

Commercial rubbers always contain low molecular weight additives. These compounds are essential in processing and in order to obtain the end-use properties of rubber. Additives may accelerate the vulcanization of rubber, can improve or modify the mechanical properties (fillers and reinforcements), modify the color and appearance (pigments, dyestuffs), give resistance to heat degradation (antioxidants and stabilizers), and provide resistance to light degradation (UV stabilizers). They can also improve the flame resistance of rubber. The identification and quantification are difficult tasks because of wide variety of different additives. Usually, mixtures of additives are used, and the added amount is often low and can be further decreased due to degradation [4].

Most analytical methods reporting the determination of rubber additives require previous extraction of the additives from the investigated material. For this purpose, widely used sample preparation techniques such as liquid extraction with various solvents, Soxhlet extraction, ultrasonic-assisted extraction, and pressurized liquid extraction (like microwave assisted extraction, accelerated solvent extraction, supercritical fluid extraction, and extraction processes employing autoclaves) are used [4]. All these methods are laborious and time-consuming. Subsequent analysis of the extracted additives has been performed using FTIR and UV-spectroscopy or by using chromatographic methods, like thin-layer chromatography, supercritical fluid chromatography, GC, high-performance liquid chromatography (HPLC), and capillary electrophoresis (CE) or their conjunction with mass spectrometry (GC/MS, HPLC/MS CE/MS). Also headspace-solid phase microextraction and thermal extraction in conjunction with GC/MS have been applied for the extraction and identification of several common rubber additives. In such volatile removal techniques, the additives are usually detected at temperatures below the decomposition temperature of the rubber [4].

Analysis of the components of vulcanized rubber is a difficult, yet very important process in analytical polymer chemistry. Each rubber compound in automobile tire contains rubber polymers, sometimes one but often a blend of two or more. The more commonly used polymers are NR, IR, BR, and SBR. SBR is generally used in the tire treads of passenger cars and NR in the relatively large-sized tires of buses and trucks [4].

The analytical Py-GC/MS has extended the range of possible tools for characterization of natural and synthetic rubbers. In the following failure analysis case, a new passenger car tire and an operationally stressed passenger car tire with cracks and

**FIGURE 17.11**

SEM image of a fracture surface of the damaged car tire, indicating the brittle nature of the component failure.

around in the grooves were investigated. [Figure 17.11](#) shows the SEM image of a fracture surface of the damaged car tire, indicating the brittle nature of the component failure. [Figure 17.12](#) shows the pyrograms of the examined rubber samples from the tire treads pyrolyzed at 700 °C. Both samples were identified as SBR. The detailed identification of the pyrolysis products of SBR was described in previous publication of the author (P. K.) [2]. In the meticulous examination of all detected compounds in [Figure 17.12a and b](#) and with the help of the mass spectral library *NIST 05*, it was found that the investigated SBR samples included the vulcanization accelerator, antioxidants, and a plasticizer but the corresponding content of the additives was not the same in both samples [10]. The identified benzothiazole ([Figure 17.13](#), RT=26.73 min) had been formed by the thermal degradation of MBT. MBT is used as an accelerator for the vulcanization of rubber and as an antioxidant. The content of the component measured by peak area ratio was two times greater in the damage tire tread when compared to the new car tire. The sulfur content from MBT increases the hardness and weakens the elastic properties of the rubber. This can lead to brittle cracking of the tire. The peak of *p*-*tert*-butylphenol was only seen in the chromatogram of the broken car tire tread ([Figure 17.13](#), RT=27.93 min). The substance was probably generated from the antioxidant BHT. The plasticizer *p*-isopropenylphenol was only detected in the broken car tire tread ([Figure 17.14](#), RT=28.13 min). The content of the antioxidant allyl cresol was about 2.5 times greater in the damaged tire tread compared to the new tire tread ([Figure 17.14](#), RT=29.43 min). However, the content of the identified *N*-phenyl-1,4-benzenediamine ([Figure 17.12](#), RT=37.48-37.49 min) was about 2.75 times

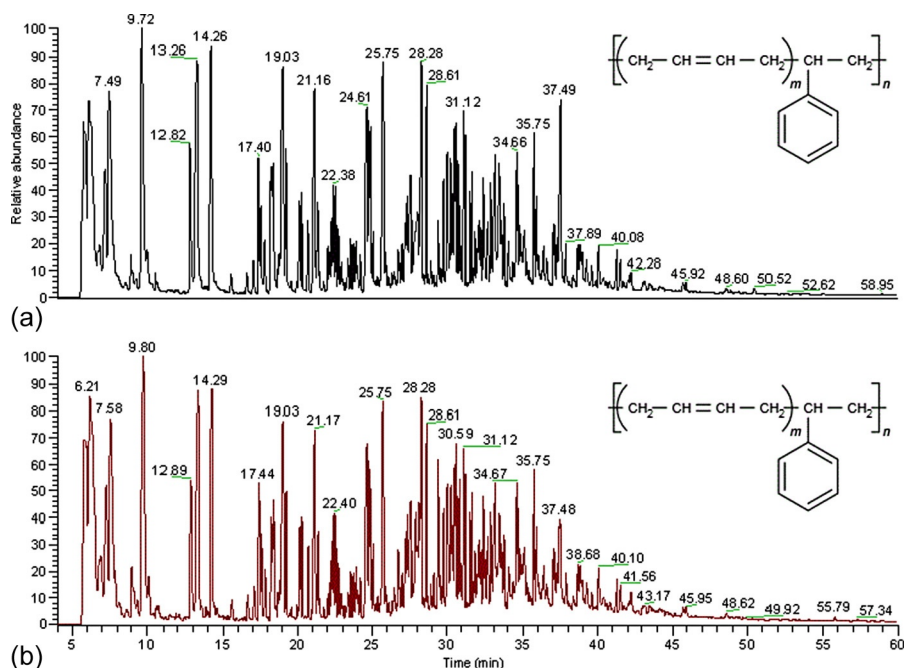


FIGURE 17.12

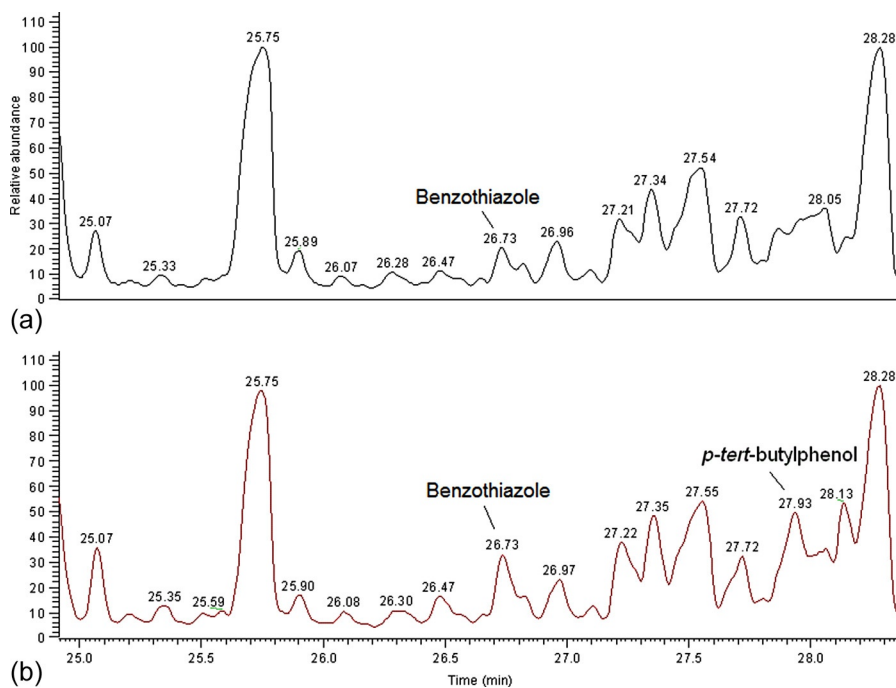
Pyrolysis-GC/MS chromatograms at 700 °C of (a) new passenger car tire tread and (b) operationally stressed passenger car tire tread. Pyrolysis-GC/MS apparatus 1 and analytical conditions for apparatus 1.

greater in the new tire tread as in the damaged tire. *N*-phenyl-1,4-benzenediamine was generated from *N*-(1-methylethyl)-*N'*-phenyl-1,4-benzenediamine during pyrolysis of SBR. This substance is a very effective antioxidant and antiozonant that provides medium- to long-term protection for all synthetic and NRs [4].

The identified differences in the content of additives in the rubbers could be the cause for the failure of the car tire.

3.3 APPLICATION OF Py-GC/MS IN FAILURE ANALYSIS IN THE CHEMICAL INDUSTRY

The goal of the next failure analysis case was to clarify why two batches of varnishes from the chemical industry show different properties. Both samples, the heated solid varnish (120 °C) and the liquid varnish, were pyrolyzed at 700 °C. The obtained pyrograms are shown in Figure 17.15a and b. Based on the decomposition products (Table 17.4), the material of both varnishes was identified as a

**FIGURE 17.13**

Pyrolysis-GC/MS chromatograms at 700 °C (range 24.9–28.4 min) of (a) new passenger car tire tread and (b) operationally stressed passenger car tire tread. Pyrolysis-GC/MS apparatus 1 and analytical conditions for apparatus 1.

mixture of water-soluble PVA and PVP polymers. To confirm the results, the standards of PVA and PVP polymers were pyrolyzed at 700 °C. Figure 17.16 shows the obtained pyrograms. The main feature of the pyrolysis of PVA is the formation of acetaldehyde, 2-butenal (crotonaldehyde), and benzaldehyde. The main decomposition product of PVP is the monomer *N*-vinyl-2-pyrrolidone (Figure 17.16). The monomer *N*-vinyl-2-pyrrolidone was produced by retropolymerization of PVP at 700 °C. The detailed identification results of pyrolysis products, PVA and PVP, are summarized in Table 17.4. Additionally, the pyrolyzate of the solid varnish showed evidence of 2-ethyl-1-hexene (RT=9.073 min) and 2-ethylhexanal (RT=11.468 min) (Figure 17.15a and b), which indicates the presence of the plasticizer bis(2-ethylhexyl) phthalate in the investigated material. The detailed thermal decomposition of bis(2-ethylhexyl) phthalate was presented in earlier publications of the author (P. K.) [2,4].

The presence of the plasticizer in one of the batch was the cause of the different properties of the varnishes.

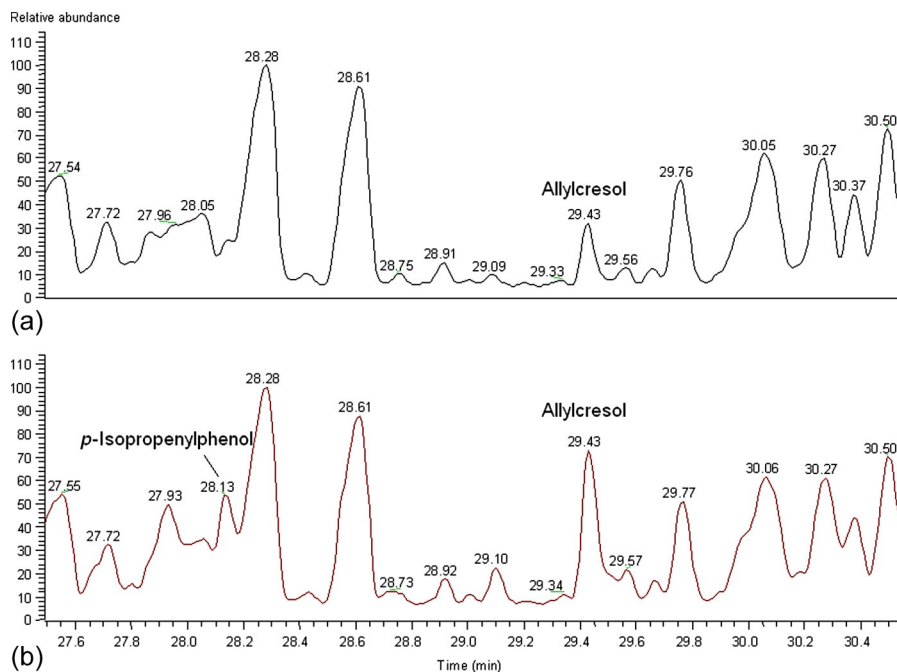
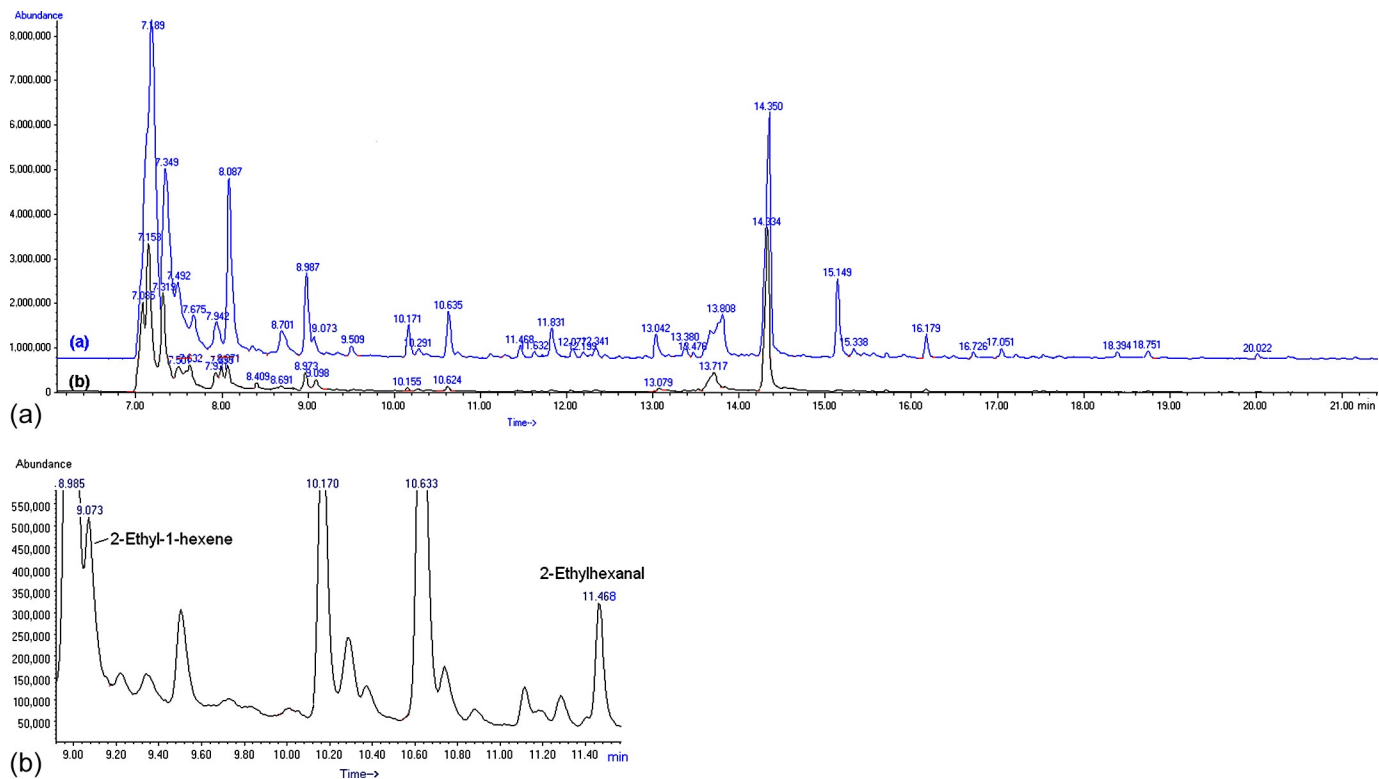


FIGURE 17.14

Pyrolysis-GC/MS chromatograms at 700 °C (range 27.5–30.55 min) of (a) new passenger car tire tread and (b) operationally stressed passenger car tire tread. Pyrolysis-GC/MS apparatus 1 and analytical conditions for apparatus 1.

3.4 APPLICATION OF Py-GC/MS IN FAILURE ANALYSIS OF A MEDICAL DEVICE

In the final failure analysis case, the cause of failure of a medical oxygen resuscitation apparatus was investigated. In the study of the components, a residue on the surface of a valve block unit from the device was observed (Figure 17.17). The residue was pyrolyzed at 700 °C. Figure 17.18 shows the pyrogram of the sample. The substances in pyrolyzate were identified using the mass spectral library *NIST 05*. Based on the thermal decomposition products showed in Figure 17.18, the examined residue was identified as a mixture of polytetrafluoroethylene (PTFE), silicone gum, polystyrene, and a large number of drugs, like oxazepam (peak 7, RT = 17.04 min), triazolam (peak 8, RT = 19.50 min), phenytoin (peak 10, RT = 26.58 min), clozapine (peak 11, RT = 27.47 min), nordazepam (peak 12, RT = 28.30 min), levomepromazine (peak 13, RT = 29.74 min), clomipramine (peak 15, RT = 34.93 min), and buprenorphine (peak 16, RT = 36.01 min). The peak of the monomer tetrafluoroethylene (peak 1, RT = 6.78 min) was assigned to PTFE, the peaks of hexamethylcyclotrisiloxane (peak 3, RT = 8.85 min) and octamethylcyclotetrasiloxane (peak 6,

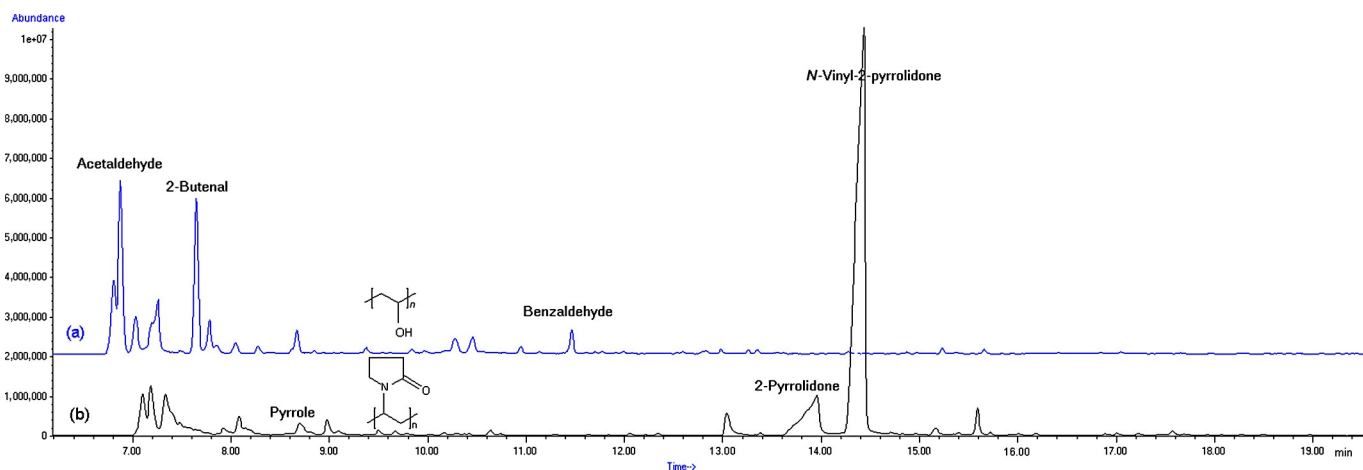
**FIGURE 17.15**

(a) Pyrolysis-GC/MS chromatograms at 700 °C of the examined (a) solid varnish (b) liquid varnish. Pyrolysis-GC/MS apparatus 2, chromatographic conditions 2. For peak identification, see [Table 17.4](#). (b) Pyrolysis-GC/MS chromatogram at 700 °C (range 8.95–11.55 min) of the examined solid varnish. Pyrolysis-GC/MS apparatus 2, chromatographic conditions 2. For peak identification, see [Table 17.4](#).

Table 17.4 Pyrolysis Products of Varnishes and References of PVP and PVA Polymers at 700 °C

Retention Time (min)	Pyrolysis Product	Heated Varnish	Liquid Varnish	PVP Reference	PVA Reference
7.04	Carbon dioxide	X	X		
7.08	Propane/ propylene	X	X	X	
7.15-7.19	Acetaldehyde	X	X		X
7.32-7.35	Acetonitrile	X	X	X	
7.63-7.67	2-Butanal (crotonaldehyde)	X	X		X
8.01-8.09	Benzene	X	X	X	X
8.70	Pyrrole	X	X	X	
8.97-8.99	Toluene	X	X	X	X
9.07	2-Ethyl-1-hexene	X			
9.51	3-Methylpyridine	X	X	X	
10.16-10.17	Ethylbenzene	X	X		X
10.29	<i>p</i> -Xylene	X	X		X
10.62-10.63	Styrene	X	X		X
11.47	2-Ethylhexanal	X			
11.63	Propylbenzene	X			X
11.83	Benzaldehyde	X	X		X
12.08	α -Methylstyrene	X	X		
13.04-13.08	1-Methyl-2-pyrrolidone	X	X	X	
13.38	Indene	X	X		
13.72-13.81	2-Pyrrolidinone	X	X	X	
14.33-14.35	<i>N</i> -Vinyl-2-pyrrolidone	X	X	X	
15.15-15.17	<i>N</i> -Acetyl-2-pyrrolidone	X	X	X	
16.16-16.18	Naphthalene	X	X		X
18.39	Methylnaphthalene	X		X	
20.02	Biphenyl	X	X		X

Pyrolysis products of the plasticizer bis(2-ethylhexyl) phthalate are indicated in grey.

**FIGURE 17.16**

Pyrolysis-GC/MS chromatograms at 700 °C of (a) poly(vinyl alcohol) (PVA) and (b) polyvinylpyrrolidone (PVP) standards. Pyrolysis-GC/MS apparatus 2, chromatographic conditions 2. For peak identification, see [Table 17.4](#).



FIGURE 17.17

Residue on the surface of a valve block unit from the failed medical oxygen resuscitation device.

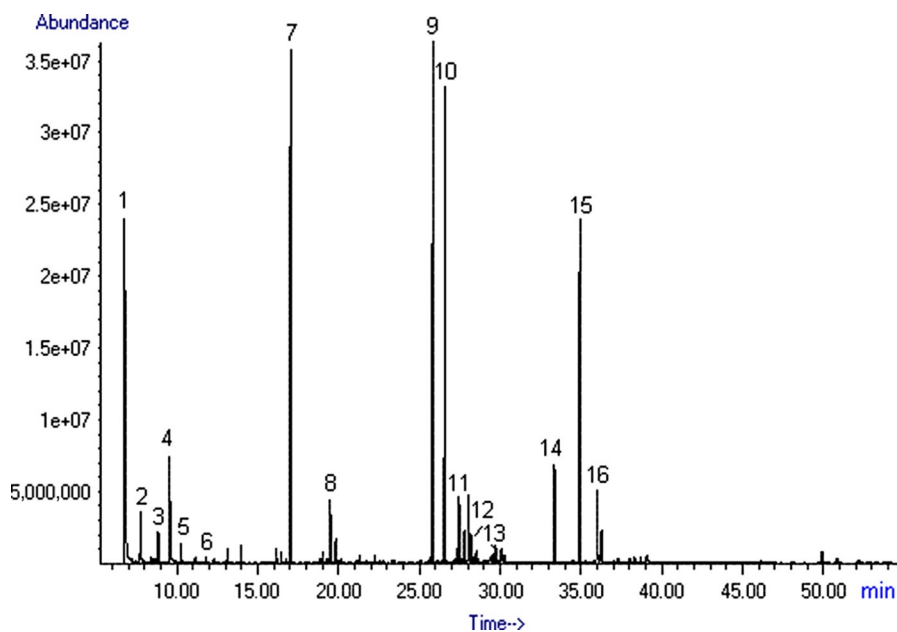


FIGURE 17.18

Pyrolysis-GC/MS chromatogram at 700 °C of the residue on the surface of the valve block unit from the failed medical oxygen resuscitation device. Pyrolysis-GC/MS apparatus 2, chromatographic conditions 2. Peak identification: 1—tetrafluoroethylene; 2—benzene; 3—hexamethylcyclotrisiloxane; 4—not identified; 5—styrene; 6—octamethylcyclotetrasiloxane; 7—oxazepam; 8—triazolam; 9—not identified; 10—phenytoin; 11—clozapine; 12—nordazepam; 13—levomepromazine; 14—not identified; 15—clomipramine; 16—buprenorphine.

RT = 11.83 min) were assigned to silicone gum, while the peaks of benzene (peak 2, RT = 7.76 min) and styrene (peak 5, RT = 10.26 min) were formed as a result of pyrolysis of polystyrene (PS).

The identified residue on the surface of the valve block unit was obviously the cause of failure of the medical oxygen resuscitation device.

4 CONCLUSION

Analytical Py-GC/MS has been proven as a valuable technique for analysis and identification of polymeric materials in the failure analysis of materials used in the chemical, rubber, and automotive industries. SEM allows the exact localization of damage of components from engineering and automotive industry. The extraordinary depth of sharpness provides deeper insights into cracking propagation and mechanical failure aspects, for example, brittleness, filler- or fiber-matrix adhesion, or stress-corrosion cracking. The used analytical techniques allow the confirmation of source of failed product, the identification of contaminants causing failure, the competitive analysis, as well as overcoming a problem in product development.

The obtained analytical results were then used for troubleshooting and remedial action of the technological process.

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