



METALLURGICAL FAILURE ANALYSIS



METALLURGICAL FAILURE ANALYSIS

Techniques and Case Studies

KANNADI PALANKEEZHE BALAN

BSP **BS Publications**
A Unit of **BSP Books Pvt. Ltd.**



Elsevier

Radarweg 29, PO Box 211, 1000 AE Amsterdam, Netherlands

The Boulevard, Langford Lane, Kidlington, Oxford OX5 1GB, United Kingdom

50 Hampshire Street, 5th Floor, Cambridge, MA 02139, United States

© 2018 BSP Books Pvt. Ltd. Published by Elsevier Inc. All rights reserved.

Distributed in India, Pakistan, Bangladesh, and Sri Lanka by BS Publications.

No part of this publication may be reproduced or transmitted in any form or by any means, electronic or mechanical, including photocopying, recording, or any information storage and retrieval system, without permission in writing from the publisher. Details on how to seek permission, further information about the Publisher's permissions policies and our arrangements with organizations such as the Copyright Clearance Center and the Copyright Licensing Agency, can be found at our website: www.elsevier.com/permissions.

This book and the individual contributions contained in it are protected under copyright by the Publisher (other than as may be noted herein).

Notices

Knowledge and best practice in this field are constantly changing. As new research and experience broaden our understanding, changes in research methods, professional practices, or medical treatment may become necessary.

Practitioners and researchers must always rely on their own experience and knowledge in evaluating and using any information, methods, compounds, or experiments described herein. In using such information or methods they should be mindful of their own safety and the safety of others, including parties for whom they have a professional responsibility.

To the fullest extent of the law, neither the Publisher nor the authors, contributors, or editors, assume any liability for any injury and/or damage to persons or property as a matter of products liability, negligence or otherwise, or from any use or operation of any methods, products, instructions, or ideas contained in the material herein.

ISBN: 978-0-12-814336-0

Library of Congress Cataloging-in-Publication Data

A catalog record for this book is available from the Library of Congress

British Library Cataloguing-in-Publication Data

A catalogue record for this book is available from the British Library

For information on all Elsevier publications visit our website at <https://www.elsevier.com/books-and-journals>



Working together
to grow libraries in
developing countries

www.elsevier.com • www.bookaid.org

Publisher: Matthew Deans

Acquisition Editor: Christina Gifford

Editorial Project Manager: Joshua Bayliss

Production Project Manager: Swapna Srinivasan

Cover Designer: Mark Rogers

Typeset by SPi Global, India

DEDICATION

To my wife Sathi Balan, and my daughters Pallavi Menon and Sunayana Menon.

PREFACE

Engineering systems are developed and manufactured to carry out certain desired functions for specific time period. When the system is not able to perform the intended function effectively, it is said to have failed. Also if the system stops functioning prematurely it is again considered as a failure of the system. In order to know what went wrong with the system a thorough and systematic analysis is required to be carried out. Such an analysis can identify the component or part of the system that caused the primary failure. Again a metallurgical investigation of the component will pin point the exact cause of the failure of that particular component.

A systematically carried out failure analysis can be useful in understanding the exact cause of failure and also the suggestions for remedial measures can help in preventing or minimizing future occurrences of such failures. The knowledge gained through case studies can help in understanding key factors that influence the properties, life, and functioning of the component. It will also help in correct design, development, and manufacture of components that will live the desired service.

Metallurgical failure analysis is a specialized field of applied metallurgy wherein there are very few publications available. This is because in most of the universities, failure analysis is not offered as a specialization but only as an optional subject in the postgraduate studies of metallurgical engineering. Most of the failure analysts who are working in this field enter it by a matter of chance like the author. They then struggle to learn the subject and the skill and gain the expertise by practicing over a long period.

Today science and technology of materials have advanced to a great extent but failures do occur. To develop an understanding of how components fail and what measures should be taken to avoid future failures, one needs references of case studies on component failures. This book is an attempt toward developing this understanding. The book will be beneficial to scientists and engineers alike who are engaged in research, design, development, production, and maintenance of metallic materials and products. Students studying for their undergraduate and postgraduate courses would also be greatly benefited by this book in developing an understanding of failure analysis and prevention.

The book introduces the subject of failure analysis and its importance to the readers. The tools and techniques used in a failure investigation are

briefly described. Mode of a failure is an important aspect of any failure analysis. This is understood by the study of fracture surface. Hence in this book a separate chapter is devoted to fracture and fractography of metals and alloys. This is followed by various case studies, which is classified according to the major causes of failures mainly from metallurgical point of view. For further reading the readers' attention is drawn to various books and articles, which are listed in the bibliography.

There are a variety of components, metals, alloys, processing, heat treatments, types of loading, applications, environments, and more often than not a combination of some of these that are to be dealt by a failure analyst. It is difficult to cover all of them in its entirety in one book. The major cause of failure is taken as the basis of classification with examples of case studies. Common alloys like steel, aluminum alloys, and copper base alloys figure in these case studies. The book is an extract of over 30 years of experience in the area of failure analysis by the author.

K.P. Balan

ACKNOWLEDGMENTS

I wish to express my profound gratitude to late Dr. D.P. Lahiri for introducing me to this challenging field of failure analysis. I am thankful to Dr. A. Venugopal Reddy for his guidance during my formative years as a practicing failure analyst. I am also grateful to him for allowing me to team up with him in the investigation of a number of case studies for nearly two decades. I am indebted to my colleagues at DMRL, Hyderabad, for their support in carrying out failure case studies. I am especially grateful to Y. Narender, who assisted me on failure analysis for nearly a decade. I also acknowledge with gratitude the support extended by N.B. Jagtap and M. Sai Madhav.

I am thankful to all my previous directors of DMRL, Hyderabad, for encouraging me to practice failure analysis as a profession. I acknowledge my sincere thanks to Dr. Amol. A. Gokhale, outstanding scientist and director, Defence Metallurgical Research Laboratory, Hyderabad, for according permission to publish the book.



Introduction

An engineering system is designed, developed, and manufactured to fulfill certain human requirements over a total technical life. In addition, the system is also expected to be reliable and cost effective. The realization of the end product goes through the following stages: conceptualizing the system for a given function, designing and development of the prototype, user trials, manufacture, and assembly. An engineering system consists of various components. The designer is required to finalize the geometry and dimension of each and every part, select suitable materials based on availability, draw a fabrication methodology, stipulate the qualitative check and acceptance norms, state the assembling technique, and finally list out the maintenance procedure and schedule. Whenever a system fails to perform a required function, any of the above stage could be the culprit. Human error caused during application and utilization of the system could be another important source responsible for failure. A failure analysis carried out scientifically and systematically alone can pinpoint the cause of such a failure. The suitable remedial measures suggested by the failure analyst can go a long way in minimizing the recurrence of such failures.

Failure of an engineering component is a condition when it no longer is able to perform the desired function due to a breakage, fracture, crack, wear, corrosion, and inaccuracies as regards to dimensional or other parameters. In a large number of engineering failures, it has been noticed that unsuccessful or nonperformance of an engineering system is due to failure of one or more component of the assembly. During preliminary examination the system's manager is generally able to locate the component that is suspected to have caused the failure. Whenever there is a crack or fracture telltale evidence is left on the failed component. In certain cases, when there is a rubbing action by the mating fracture surfaces or rubbing of the fracture surfaces by some other component it amounts to loss of useful evidences for an analyst. In such a situation, a failure analyst has to resort to the available data and circumstantial evidences to draw a conclusion about the failure. A good failure analysis can be highly useful in understanding the cause of failure and also the methods to prevent occurrence of such failures in future.

Failure analysis is a combination of an art and a science. All material scientists or engineers cannot handle failure analysis as it needs certain amount of experience and skill to handle a case study. Since every failure is reported as a crisis it is also expected to get an investigation carried out within a very short time. However, it should be borne in mind that a failure analysis cannot be performed with a magical wand. The analyst requires a certain amount of time to understand the background, collect suitable samples, make a variety of specimens for a large number of tests, interpret the test results, and then come to a conclusion in order to pronounce the cause. Failure analysis is based on sound metallurgical principles and authentic scientific evidences gathered through standard tests. The time frame demanded by the analyst should be considered to be genuine and must be provided for a job to be well handled. After all everyone involved in the case looks forward to the cause of failure and a suitable remedial measure in order to avoid future recurrence of the same.

Abuse of metallurgy has been the primary reason for a large number of failures of metallic components in our country. Most of the small- and medium-scale industries appoint mechanical engineers in all such fields of shop floor where a metallurgist should actually be taking care of the job, for example, a heat treatment shop. During the design and development of a system, for example, a weapon, ordinance, or an aerospace vehicle; metallurgist or materials experts are seldom part of the team. When some component of the system fails, which is bound to happen, the development team appoints an inquiry committee to look into the matter. Even at this stage a metallurgist is not appointed as one of the members in the inquiry committee. The engineers of the development team arrive at a cause of failure based on the operational error or breakage of some part of the system. Now they approach a metallurgical laboratory with a request for chemical analysis of the failed component. Sometimes the chemical analysis is also accompanied with a request for hardness check. The inquiry committee thinks that it knows the cause of failure and what more they want is a confirmation by a chemical analysis or hardness check. Even at this stage the committee does not realize that what they require is a failure analysis by an expert. What they do not realize is that they need to approach a materials consultant for a thorough failure analysis and appointment of a metallurgist for all future endeavors. This nonrealization leads to failure after failure but fails to wake them up to understand their basic fault—‘abuse of metallurgy’. The day this realization comes and metallurgists get the right place in

engineering industries and engineering system development team, the problem of premature failure would minimize to a large extent. Whenever an engineering component fails it is advisable to approach a failure analyst for a detailed failure investigation and apply the remedial measures suggested by him, which will help in avoidance of failures in future.



Fracture

When a metallic body is subjected to a stress it withstands the applied stress, which can be understood by a typical tensile test. When the body is no longer able to withstand the applied stress it starts cracking. Cracking goes through two stages viz. crack initiation and crack propagation. Formation of crack, due to release of stress in a metallic body that ultimately results in separation of the body into two or more parts, is known as fracture. A majority of component failures is due to fracture and therefore an understanding of this phenomenon forms an important aspect of failure analysis. There are different types of fracture viz. ductile, brittle, fatigue, etc. Each type leaves certain characteristic fracture features on the fractured surface that lead us to know the mode of failure. Knowledge of the mode of fracture helps us in working toward the cause of failure.



2.1 DUCTILE AND BRITTLE FRACTURES

Fractures can be classified broadly as ductile fractures and brittle fractures. A ductile fracture is associated with considerable plastic deformation. An appreciable amount of gross plastic deformation in the form of change in dimension (increased length or local decrease in cross section) can actually be observed in the case of a ductile fracture. A typical cup and cone fracture of a ductile metal as observed in a tensile test is an example of a ductile fracture (Fig. 2.1). An appreciable amount of shear lip as seen in Fig. 2.1 is an evidence of a ductile fracture. A ductile fracture is characterized by a fibrous fracture surface with a prominent shear lip. Thus a typical cup and cone fracture of a moderately ductile metal, a rupture following drawing of a highly ductile metal viz. gold or lead to a point and mere separation of a metal by 45° shear are all examples of ductile fractures.

A brittle fracture is characterized by very little micro-deformation and negligible gross deformation. In a brittle fracture the rapid rate of crack propagation ultimately results in fragmentation. A coarse and crystalline

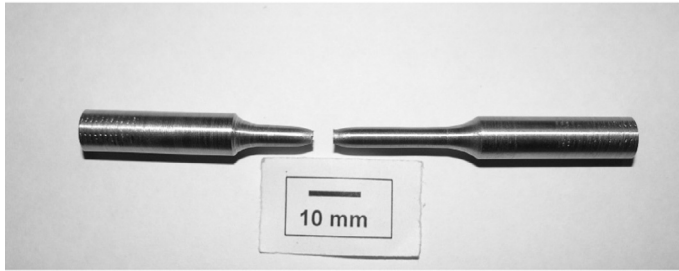


Fig. 2.1 A typical cup and cone fracture of a ductile metal as observed in a tensile test.

fracture surface is an evidence of a brittle fracture. An otherwise ductile metal can fail in a brittle mode due to three main factors viz.

- (i) a triaxial state of stress,
- (ii) a low temperature, and
- (iii) a high strain rate.

Besides the influence of the three physical factors, as mentioned earlier, there are many metallurgical factors and a few environmental factors, which are responsible for the transition of ductile to brittle fracture in a metallic system. The embrittlement of metals is mainly due to the influence of strain, temperature, time, and environmental conditions. The various types of embrittlement of metals are temper embrittlement, tempered martensite embrittlement, hydrogen embrittlement, etc.



2.2 FRACTOGRAPHY OF DUCTILE AND BRITTLE FRACTURES

The fracture surface of a ductile fracture shows a dull and fibrous appearance when examined visually or under a stereomicroscope. Fracture surface of a ductile fracture when examined under a scanning electron microscope (SEM) reveal, what are called, 'dimples'. Dimples look like a collection of nests of weaver birds when viewed perpendicular in to the nest. The scooped out region invariably shows undissolved or precipitated particles from where the voids had originated. Dimples are formed by the initiation and coalescence of micro-voids around particles. With continued deformation the voids increase in size and eventually the interconnecting ligaments of the material get separated resulting in dimples. The density, size, shape, and depth of dimples observed on the fracture surface

correspond to the extent of plastic deformation the metal has undergone prior to fracture. Highly ductile materials reveal small numbers of large-sized and deep dimples. Materials with relatively lower ductility show larger number of finer and shallower dimples. Such evidences suggest that the amount of plastic deformation required for the process of separation is relatively less. On the flat areas of ductile fracture equiaxed dimples are observed whereas on the shear lips, highly deformed zones, in bending and shear fractures, elongated dimples are observed. Typical fractographs showing equiaxed and elongated dimples are illustrated in Figs. 2.2 and 2.3, respectively.

Brittle fractures have a bright and crystalline appearance when observed visually or under stereomicroscope. The fracture surface invariably shows, what are chevron marks pointing toward the origin of the fracture. Chevron marks are also called herringbone pattern as shown in Fig. 2.4. Fracture surface of a brittle fracture when examined under SEM reveals a transgranular cleavage or an intergranular fracture. A transgranular cleavage is formed by the propagation of cracks along well-defined crystallographic planes. Numerous cleavage-crack segments look like tributaries, which ultimately join the main

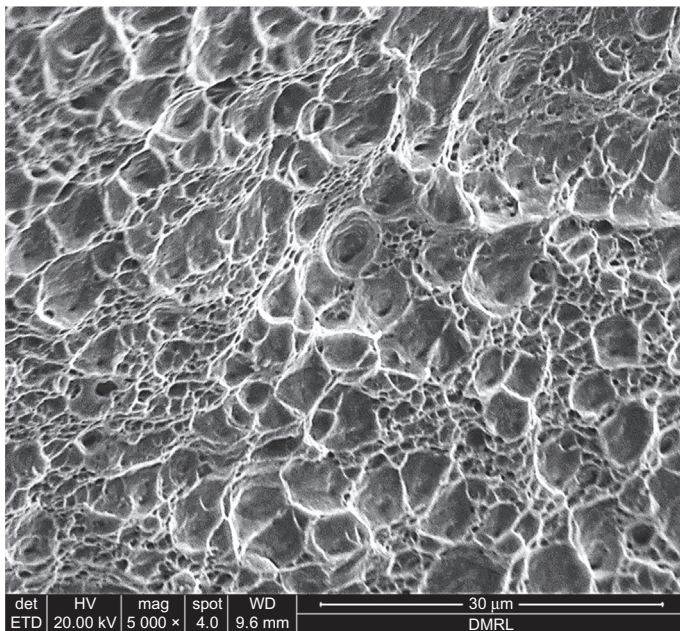


Fig. 2.2 Typical fractograph showing equiaxed dimples.

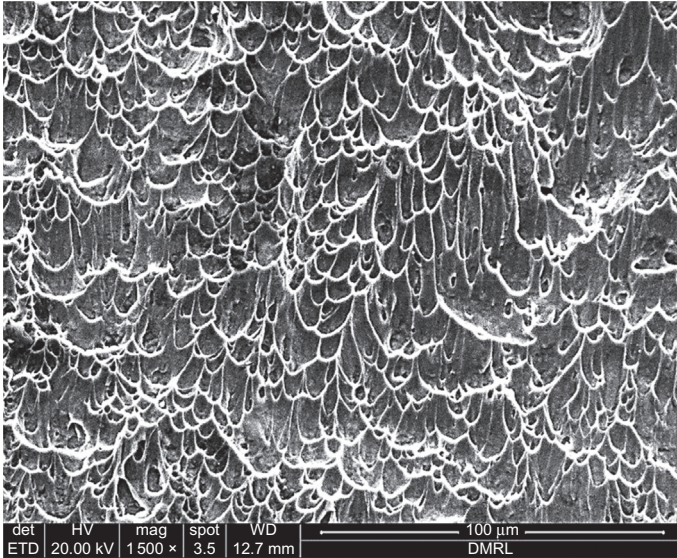


Fig. 2.3 Typical fractograph showing elongated dimples.

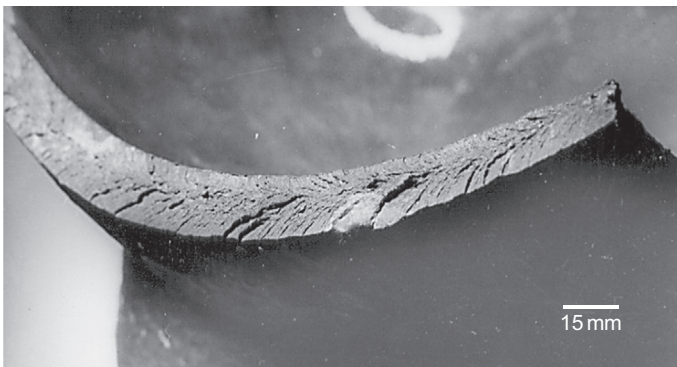


Fig. 2.4 Fracture surface showing chevron marks or herringbone pattern.

crack, forming what is called a “river pattern”. A typical transgranular cleavage fracture, which shows the direction of crack propagation, is shown in [Fig. 2.5](#). An inherently brittle metal or embrittlement of a ductile metal due to the influence of certain physical factors, as mentioned earlier, reveals transgranular cleavage fracture features. Fractures revealing intergranular fracture features are another kind of brittle fracture, which develop only under certain metallurgical factors influencing embrittlement of the metal. Intergranular fracture

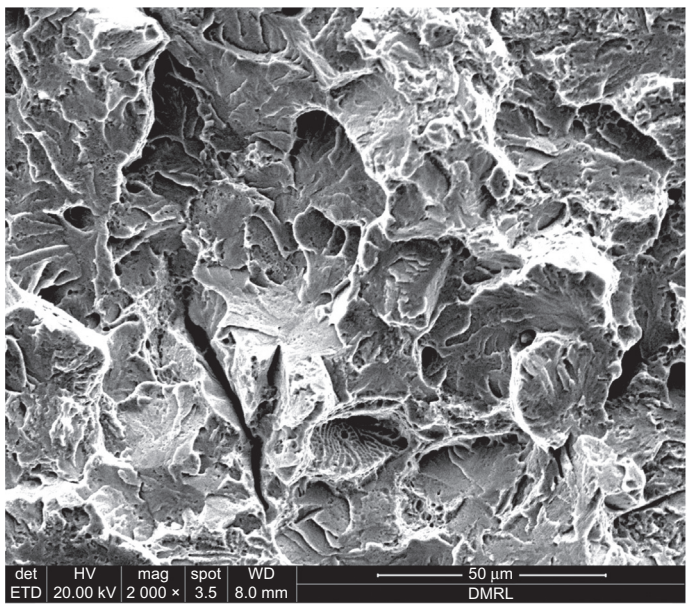


Fig. 2.5 SEM fractograph showing a typical transgranular cleavage fracture.

reveals polygonal grain facets and at times secondary intergranular cracks. A typical fracture feature of an intergranular fracture is shown in [Fig. 2.6](#). Due to metallurgical or environmental factors when the grain boundary becomes a weak path for crack propagation an intergranular fracture results. Impurity

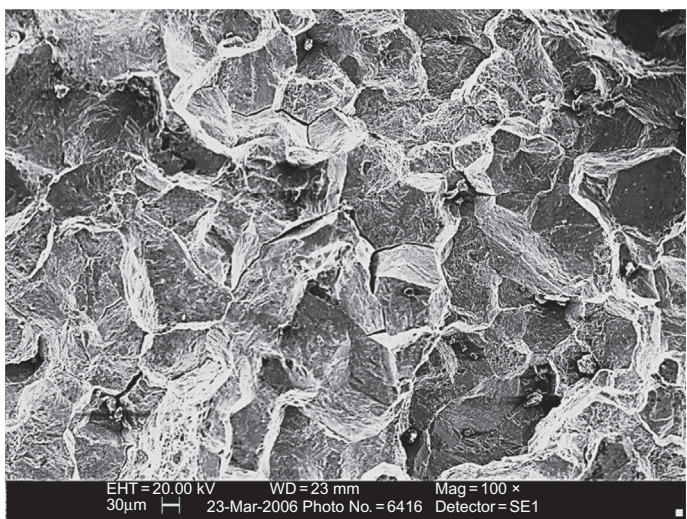


Fig. 2.6 SEM fractograph showing a typical intergranular fracture.

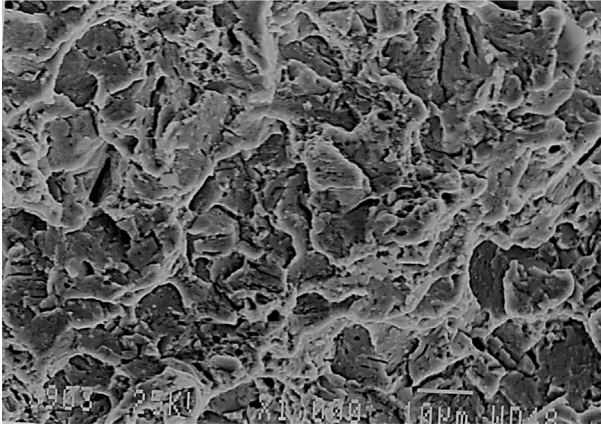


Fig. 2.7 SEM fractograph showing a typical quasi-cleavage fracture.

segregation, formation of cavities and cracks due to high temperature exposure, and corrosion of grain boundaries are all reasons for a metal to lose its cohesive strength and fail in an intergranular mode.

Certain metals, which do not show crack propagation along well-defined crystallographic planes, end up in what is called a “quasi-cleavage” fracture. The fracture surface would reveal some amount of ductile fracture feature or dimples along with cleavage fracture. A typical quasi-cleavage fracture surface is shown in [Fig. 2.7](#). In case of a ductile to brittle transition due to physical factors a quasi-cleavage fracture is revealed.



2.3 FATIGUE FRACTURE

As the fracture mode in case of fatigue is different from that observed in case of a ductile overload fracture or a brittle fracture, it necessitates a special mention under this topic on fracture. Failures occurring under conditions of cyclic loading are called fatigue failures. Even though a ductile metal like aluminum or stainless steel fails by fatigue it shows a brittle fracture with no gross deformation near the fracture. Fatigue crack propagates in a direction perpendicular to the principal tensile stress. Fatigue fracture is characterized by a smooth fracture surface that generally extends by one-half to two-thirds of the cross section followed by a fibrous fracture zone. The smooth fracture region depicts the section in which fatigue crack had propagated during the course of cyclic loading of the component. The region is seen to be smooth due to the rubbing action of the mating surfaces during the stage of fatigue crack propagation. In certain cases of fatigue failures, the

smooth region of the fracture surface would reveal a number of ripple markings, which are called “beach marks” or “clamshell” marks. The progress of fracture would generally be seen originating from a lens shaped or crescent like area from where the beach marks will be found diverging. Beach marks occur as a result of change in loading or by oxidation of the cracking surface during the interval between stop and start or reversal of loading of the part. A typical fatigue fracture surface is shown in Fig. 2.8. There may be situations where the fatigue crack growth is uninterrupted. In such cases fatigue fractures do not exhibit beach marks. The final fracture zone that shows a fibrous fracture surface usually culminating in a shear lip is due to ductile overload as the remaining cross-sectional area is no longer capable of withstanding the applied load. Fatigue fracture can sometimes exhibit multiple origins as shown in Fig. 2.9.

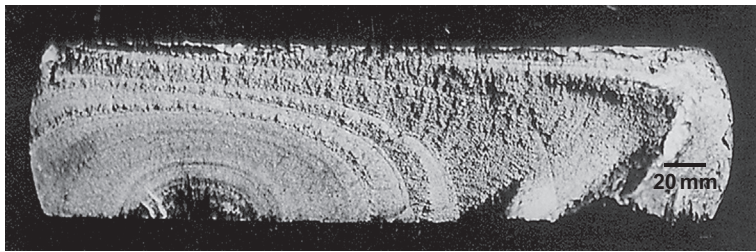


Fig. 2.8 SEM fractograph of a typical fatigue fracture surface showing beach marks originating from a lens shaped dark region.

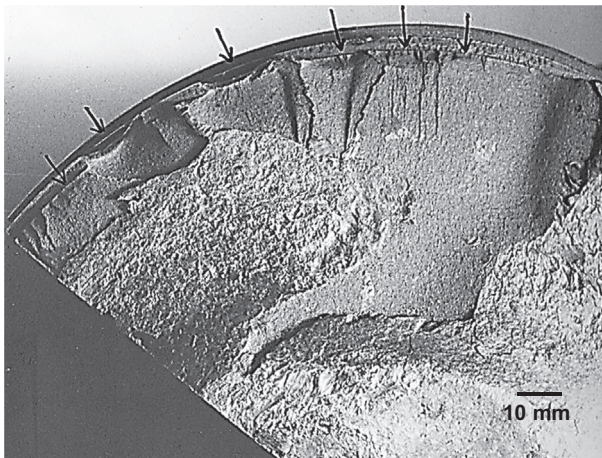


Fig. 2.9 Fatigue fracture surface showing multiple origin (arrow heads).



2.4 FRACTOGRAPHY OF FATIGUE FRACTURE

On examination under SEM the smooth region of a fatigue fracture invariably shows finely spaced lines called 'fatigue striations'. The fatigue striations are oriented perpendicular to the direction of crack propagation (Fig. 2.10). Each striation represents one single load excursion. Well-defined fatigue striations are revealed in ductile metals while it may be difficult to observe fatigue striations in high strength metals and alloys. Fatigue striations are usually observed at high magnification generally at 1000X or more under SEM. Materials having lower ductility show brittle or broken striations (Fig. 2.11). In case of corrosion fatigue or high temperature fatigue the fatigue striations may be indistinct or obscured by corrosion products due to the effect of corrosive environment or high temperature, which are associated with the cyclic loading conditions of the component (Figs. 2.12 and 2.13).

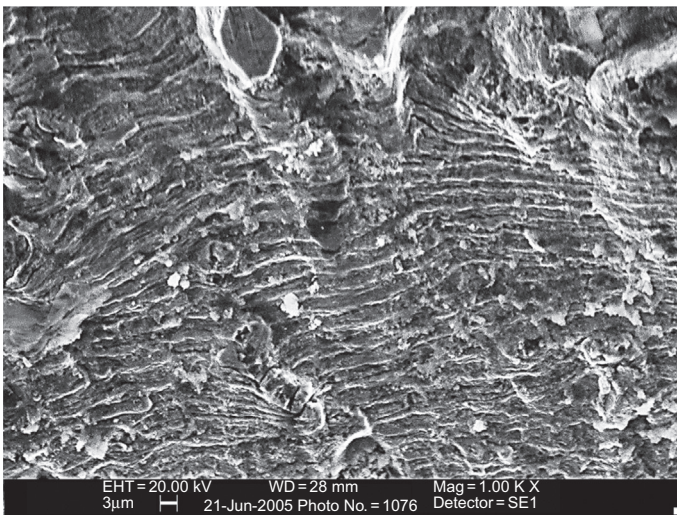


Fig. 2.10 SEM fractograph showing well-defined fatigue striations in ductile metals.

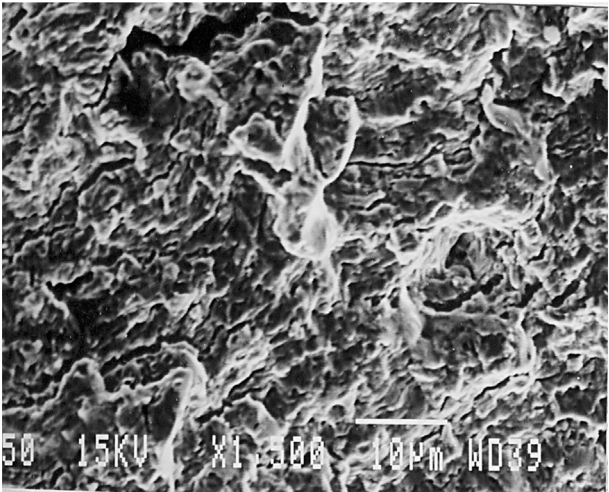


Fig. 2.11 SEM fractograph showing ill-defined fatigue striations in brittle metals.

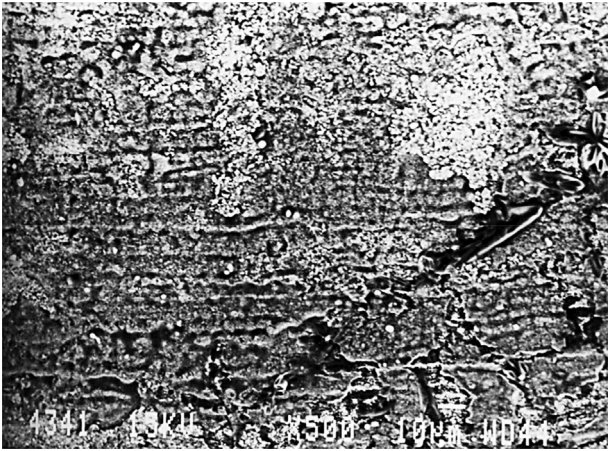


Fig. 2.12 SEM fractograph showing striations with corrosion products.

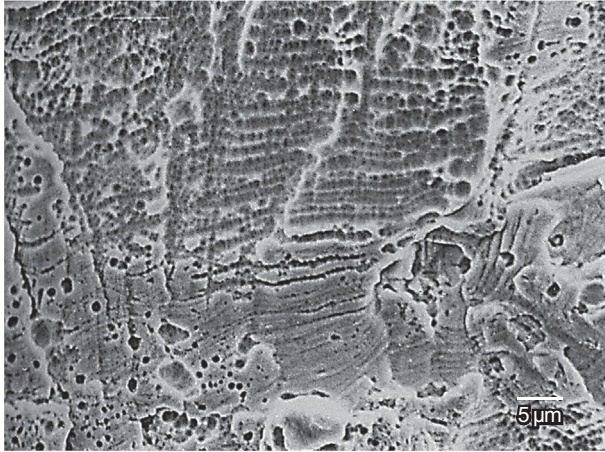


Fig. 2.13 SEM fractograph showing striations with porosity.



Tools and Techniques

Depending upon the nature of a failure, a failure analyst chooses a particular direction of investigation. Nonetheless the following broadly comprise a failure analysis. Each stage, as described later, requires certain tools and techniques, which is utilized to its best by the analyst.



3.1 ACQUISITION OF BACKGROUND DATA

When a failed component is received for analysis, the analyst must jot down all available data pertaining to the material, manufacturing route, mechanical properties, loads applied and conditions of loading, service conditions, possible human error, expected service life, and life at failure of the component. The information so acquired would enable the analyst to relate the failed component's intended performance with those that was expected out of it. The information on the life of the component would help in knowing whether the failure was premature or the failed component had completed a good amount of service life before failure. The data would enable the analyst to understand what was specified or stipulated and what was used or followed.



3.2 SELECTION OF SAMPLES

The best situation for the failure analyst would be to visit the accident site, understand the failure by talking to the operators and users of the system, and select the suspect components for failure analysis. Whenever the failed component is delivered to the analyst in his laboratory, it requires his skill to choose the right parts for investigation. This would avoid wastage of his time and efforts. The samples selected for analysis should represent the main characteristic of the failure. In case sufficient background data is not made available to the analyst it becomes important that a similar serviceable component be also taken for a comparative study. If the component is big enough in size and the failed portion is localized, then it is advisable to take a sample from a region, which has experienced a similar service condition but is remote from the failed region for comparison during analysis.



3.3 PHOTOGRAPHIC RECORDS

Possible photographic records must be made of the site of the accident with the system, which was involved in the accident. Then the photograph of the suspect component must be taken and recorded. In case the analyst receives the component for failure analysis, the photograph of the failed component must be recorded in the as-received condition. While taking photograph, care must be taken to record different views viz. front view, top view, etc. of the failed component. Close up view of the failed or defective location and that of the fracture surface are also to be recorded for future reference. Failure analysis being a destructive investigative method, reconstruction of the failed component in the as-received condition could be a difficult task in the absence of initial photographic records. In modern days, digital camera has come to be very useful in recording a large number of photographs of the failed components in different angles out of which the most representative ones which detail the characteristic of the failure may be utilized while reporting the failure analysis.



3.4 PRELIMINARY EXAMINATION

The failed component along with disintegrated parts, if any, must be carefully examined before undertaking any kind of cleaning. The soil, corrosion debris, paint or plating, etc. found on the failed parts or the fracture surface can provide vital clues to reconstruct the sequence of events that lead to the failure. For example, corrosion debris can give an idea about the service environment to which the component was subjected to while presence of a plating material on the fracture surface suggests that the crack existed prior to the plating stage during manufacture of the component. All such evidences gathered during preliminary examination should be noted and documented.

Preliminary examination can be carried out visually and using a hand held low power magnifying glass. Visual examination gives a fairly good idea about the method of manufacture viz. cast, forged, rolled, machined from a bar stock, etc. The surface finishes, plating, the material being magnetic or not, etc. are also known from such an examination. The analyst must take dimensional recordings of the component and document the same as it plays a vital role in the analysis. Wherever necessary, making a sketch of the part is a good practice.

Visual examination can further be carried out using a low-power microscope (stereo microscope) wherein magnifications from $5\times$ to $50\times$ helps in observing fine details in the defective part. Examination of fracture surface using stereomicroscope is highly useful in locating the origin of failure. The characteristic feature of the fracture surface viz. fibrous, crystalline, smooth fracture, beach marks, chevron pattern, etc. can be identified by stereomicroscopic examination.



3.5 NONDESTRUCTIVE TESTING

Nondestructive testing is extremely useful in detecting surface cracks, subsurface flaws, and discontinuities in a metal. Cracks and discontinuities that are open to the surface can be detected by dye-penetrant test. The method is suitable for both magnetic as well as nonmagnetic materials. For surface or subsurface flaws in ferromagnetic materials, magnetic particle inspection can be utilized. For detecting surface and shallow defects on electrical conductors, an eddy current test can be resorted to. Sound waves of very high frequency, which can be transmitted through metallic materials, are utilized in the ultrasonic inspection technique. The methods not only give flaw position within the metal but also give the flaw size. Radiography is the technique of NDT wherein X-rays or gamma rays are used in detecting flaws in metals. Flaw size and location can, not only be detected by radiography; the radiograph enables the analyst to keep a paramount record of the defect. Radiography is highly useful for evaluation of weld, casting, and corrosion related damages.

It must be borne in mind that each technique of NDT has its own limitations viz., limiting size of the defect that can be detected, type of material that will respond to the test, and depth at which the defect may be located. The failure analyst has to approximately choose the type of the technique depending on the failure case. Any standard book on nondestructive testing methods or books which familiarizes with the techniques of NDT can be referred for obtaining more details on the NDT techniques by the reader.



3.6 CHEMICAL ANALYSIS

Chemical composition of the failed component bears its significance in knowing whether the selection of the material was as per the specification or not. In cases where the user is not in a position to give the specified

composition of the failed part, chemical analysis of the part can be carried out and compared with an unused/un-failed part intended for similar application. In a number of case studies chemical analysis makes the report complete.

Chemical analysis can be carried out both by 'wet analysis' as well as 'instrumental analysis' methods. Depending upon the element to be analyzed the right method is chosen, for example carbon and sulfur analyzer, whereas chromium or nickel in steels are analyzed by wet analysis. It is better if qualitative/semiquantitative analysis is first conducted using a spectrographic or X-ray fluorescence analysis, which gives the nature of major, minor, and trace elements and their approximate quantity after which the right technique for quantitative analysis can be decided. This is basically carried out as most of the analytical techniques utilize comparison methods using standards. The standards having the right ranges within which the composition of the test sample lies have to be chosen for a chemical analysis. Therefore, it is beneficial to have an idea of the nominal composition of the failed component before the detailed chemical analysis is carried out.

Another important aspect of chemical analysis is the sample preparation. Samples used for different techniques of analysis are different. For example, drillings, shaping, or millings, which can be easily dissolved in a chemical solvent, are used for wet analysis. Small pieces of about 3 mm × 3 mm × 3 mm are used for carbon and sulfur estimation. A flat sample of approximately 25 mm square and 5 mm thickness is required for sparking in a spectrographic analytical instrument. Similarly, the specimen sizes and shapes are different for gaseous composition analysis in a metallic material. It is advisable to consult the chemist who will be carrying out the analysis before the samples are made for any chemical analysis. It must be remembered that before the samples are cut or drilled from a part of the component, the part must be stripped of any surface plating by chemical or mechanical means and cleaned in a degreasing agent like carbon tetra chloride or acetone. Any foreign material sticking to the sample can vary the result significantly as most of the methods used require initial known weight of the sample to be taken for estimation of the weight percentage of the element in the given alloy. One has to be careful during sample preparation for carbon analysis in metals, which are carburized intentionally. In such cases the carbon enrichment on the surface up to a certain depth will be significantly higher as compared to the substrate metal. The surface layer rich in carbon has to be sufficiently ground or machined before collecting the sample representing the substrate material.



3.7 FRACTOGRAPHY, ENERGY DISPERSIVE ANALYSIS BY X-RAY (EDAX), AND MICROSCOPY USING SCANNING ELECTRON MICROSCOPE (SEM)

3.7.1 Principle of SEM

An electron beam emitted from an electron gun is focused by a multiple stage electromagnetic lens system and scanned over the specimen surface. Electron beam—sample interaction produces secondary electrons (SE), back-scattered electrons (BSE), and X-rays along with other signals, which are collected by their respective detectors and are converted to video signals. These signals are displayed on the cathode ray tube (CRT) and magnified electronically in synchronization with the scanning beam in the electron optical column. SEM has the dual advantage of a high resolution ($\approx 50 \text{ \AA}$) and a good depth of focus. These properties enable the failure analysis through a three-dimensional imaging.

3.7.2 Fractography

Secondary electron image is used for fracture surface examination. The topographical study of a fracture surface is called fractography. The characteristic feature of the fracture viz., cleavage, quasi-cleavage, intergranular, dimples or fatigue striations, etc. are observed and the mode of failure is ascertained by fractography.

3.7.3 EDAX

When an electron beam bombards a sample along with a variety of other emissions X-rays are also emitted. The EDAX unit attached to SEM collects all emitted X-rays of all energies and sorts them electronically to provide an X-ray energy spectrum. Based on the characteristic energy of elements, they are identified on the spectrum. EDAX analysis of a point or small area of interest on the fracture surface can be carried out to know the micro-chemical composition of the substance present in the localized region. The technique is highly useful to get a qualitative analysis of slag or other nonmetallic inclusions, corrosion debris, and foreign particles deposited or embedded in a sample.

The fracture surface for fractography should be degreased and cleaned ultrasonically in acetone. In case of oxide or rust on the fracture surface, a dilute solution of oxide remover can be used for removing the oxide layer by agitating the sample in the solvent ultrasonically. Cleaning with acetone

can follow this. Cleaning can also be effectively done using a neutral soap solution and water. Care must be taken not to scrub the fracture surface by hand or a scrubber during cleaning lest there is a risk of damaging the fracture surface resulting in loss of features.

3.7.4 Microscopy

The specimen for microscopic examination is prepared in a similar way as that used for optical microscopy (metallography), but the etching is carried for a little longer duration so as to obtain a deeper etching. The requirement of deeper etching is basically to create a topographical variation on the polished surface, which can produce a better image under a SEM. The microscopic examination, which requires studies at higher magnification of the order $>1000\times$ is usually carried out in SEM. For example, the study of size and shape of gamma prime (γ') in a gamma (γ) matrix in nickel base super alloys can best be carried out by SEM microscopy. Similarly, grain boundary precipitation of chromium carbide in a sensitized austenitic stainless steel can be unambiguously confirmed by SEM examination at suitable magnification.

Very useful information may be obtained from polished specimens when the imaging is done using back-scattered electrons. Polished and un-etched specimens are used for BSE imaging. Regions of high atomic number back scatter electrons more strongly than those of low atomic number and so appear brighter.



3.8 OPTICAL MICROSCOPY

The most important tool for a failure analysis is an optical microscope. Presence of defects like voids, porosities, cavities, and secondary or pre-existing cracks can be detected by optical microscopy. The identification and rating of different types of nonmetallic inclusions can be carried out by the examination of as polished specimens of metals under an optical microscope. On etching the polished specimen using a suitable etchant the microstructure is revealed when viewed under a microscope. Microscopic examination informs whether the material is composed of the correct microstructure. Microstructure indicates whether a material was manufactured by casting or was it a wrought product. The orientation of grains or second phase particles also suggests whether the wrought product was forged, rolled, extruded, or machined from a bar stock. Segregation of a phase, banding and grain boundary deposition, etc. is also identified by

optical microscopy. Optical microscopy is the right method to know whether the component was properly heat-treated or not. In fact, it informs the complete history of processing and heat treatment of a component. The case studies that are discussed in this book give a variety of investigation using optical microscopy which have been decisive in arriving at the cause of failure.

Specimens for metallography are sectioned from near the failure and also either from an area remote from failure or from a serviceable component if available. It is advisable to use longitudinal section of the sample for metallography. In case it is difficult to judge which is the longitudinal direction from the profile of a component, then two perpendicular sections are taken for metallography. The specimens are mounted by hot- or cold-mounting technique for better grip and ease of polishing. The mounted specimens are polished mechanically by standard polishing techniques to a good surface finish.



3.9 ELECTRON PROBE MICRO ANALYSIS (EPMA)

It is a modern technique used to chemically analyze various micro features present in a material. When a material is bombarded with an electron beam it knocks out electrons from the inner orbital of the atoms comprising the material. The vacancy in the core-shell thus created gets filled when an electron from one of the outer orbital falls into the vacancy. The difference in energy between the core-shell vacancy and the outer shell electron is emitted out as an X-ray photon. The atom can be identified by the energy of this X-ray photon, which gives a qualitative picture of the composition. The intensity of the X-ray gives the quantitative chemical composition of the microstructural phase.

The major studies conducted using an EPMA is qualitative and quantitative analysis of micro features, X-ray elemental dot mapping, and line scan. Since the lateral and depth resolution of EPMA is of the order of a micron, chemistry of a region of one micron can be analyzed using an EPMA.

In a failure investigation EPMA is used to determine:

- (a) Micrometric level concentration gradients at phase boundaries, grain boundaries, surfaces, diffusion couple interfaces, and welded sections,
- (b) In the determination of segregations within a single phase,
- (c) In the identification of phases and inclusions, and
- (d) In the characterization of coatings.

Mounted and polished specimens used for metallography can be used for EPMA. The surface should be either used in un-etched condition or should be very lightly etched.



3.10 X-RAY DIFFRACTION

In contrast with chemical analysis for elemental composition of the alloy, phase analysis by X-ray diffraction technique gives the presence of the actual phases or compounds. Phase analysis is based on the phenomenon of diffraction of X-rays by the crystals of the substance governed by Bragg's law: $\lambda = 2d \sin \theta$, where λ is the wavelength of the radiation, d is the inter-planar spacing of the diffracting crystal, and θ is one half of Bragg angle. An X-ray diffraction experiment provides a set of d spacing, which is matched with those of known phases in ASTM or JCPDS card files for the identification of the phase.

The X-ray diffraction technique in failure analysis is used to determine:

- (a) The residual stresses in a stressed component,
- (b) The volume fraction of retained austenite in steels, and
- (c) The qualitative nature of a substance that was deposited due to environmental factors on the component.



3.11 EVALUATION OF MECHANICAL PROPERTIES

3.11.1 Hardness

It is a simple and quick assessment of strength of a metallic material. Hardness can be reasonably related to ultimate tensile strength in case of steels. Depending upon the manufacturing process viz. cast or wrought and also the size and section thickness, one has to choose the appropriate method for measurement of hardness. For most of the failure investigation it is advisable that hardness measurement be carried out on the specimen prepared and used for metallography. In case the surface of the specimen is too dark due to etching effect, the same can be removed before hardness measurement. Vickers hardness is suitable for most of the failure analysis although for cast products Brinell hardness is advisable. In case the specification stipulates Rockwell hardness either the same can be measured or reported after suitable conversion from Vickers scale. Depending upon section thickness and hardness level suitable load can be chosen for measurement of hardness. At least three concordant values or an average of five measurements that do not vary within ± 5 HV should be reported.

Hardness of surface coating, plating, carburized and hardened layer, nitrided layer, decarburized layer, soft or hard second phases in a microstructure, etc. are measured using micro hardness tester. Depending upon the area/width of the layer or phase of interest hardness is measured either by Vickers or Knoop scale. Whenever the section thickness of a sample is too small for bulk hardness measurement, one can opt for micro hardness provided the measured value is only treated as approximate and used only for guidance.

The hardness test is carried out to evaluate the condition of the component during use or failure. The change of microstructure during service due to service conditions can be easily detected by hardness check and comparison with that of specified hardness during assembly. The improper heat treatment imparted to a component before use can also be evaluated by hardness measurement. Usually microstructural deficiency is substantiated by hardness values in a failure analysis. Increase in strength of an alloy due to aging or precipitation hardening that is difficult to ascertain by metallography can be easily confirmed by hardness measurement. Similarly work hardening due to mechanical working of metals can be substantiated by hardness check. Decarburization, formation of retained austenite and martensite, etc. due to overheating are all detected by micro hardness measurement.

3.11.2 Tensile Properties

Tensile properties are rarely evaluated by carrying out tensile test in a failure analysis. Invariably the sample received for failure analysis is found to be inadequate for machining standard test specimens for a proper tensile test. The expensive tensile test can also be avoided as the hardness value gives a fairly good idea about the strength of a metal especially in case of steels. In certain cases, when the heat treatment is found to be faulty and for confirmation of the same a heat treatment experiment is carried out on the failed component, then a tensile test can be carried out to evaluate mechanical properties in order to confirm the correctness of the modification of heat treatment. Case studies pertaining to improper selection of materials and improper heat treatment generally call for a corroboration of the findings with a tensile test. Breaking load specified for single wires in a wire rope requires a tensile test check for quality of the wire used against its specification. Whenever tensile ductility (percentage elongation and percentage reduction area) become important parameters for defining the quality of a product, carrying out a tensile test to check for the same becomes most important of all the tests.

3.11.3 Impact Toughness

Impact toughness at both room temperature and subzero temperature is an important material property in a large number of components for engineering applications. Failures occurring due to poor impact toughness generally show a brittle fracture and unacceptable microstructure. The poor impact toughness could be due to improper heat treatment which cause embrittlement of the metal viz. temper embrittlement, tempered martensite embrittlement, strain-age and quench-age embrittlement, etc. Impact test carried out on the sample from the failed component indicates the low energy absorbed by the metal. The failure analyst can establish heat treatment experiments carried out in the laboratory to de-embrittle the component in case of reversible embrittlement or for suggesting modifications in heat treatment only with an impact test. The most popular impact test is the Charpy V-notch impact test unless otherwise stipulated by the manufacturers or the users.



Material Selections

There is no generalization that is valid for material selection problems. Each problem has to be taken up individually and assessed against certain guidelines useful for selection of materials in relation to possible mechanism of failure, types of loading and stress, operating temperature, and environment. A materials consultant could be highly useful to the designer in the initial stage of component design. Failures caused due to deficiencies in material selection are often a result of absence of a materials' expert in the design team. Proper selection of materials for an intended purpose calls not only for the consideration of material characteristics but also factors such as availability, ease of fabrication, service life, and cost. Chemical composition and mechanical properties make the basic factors of material characteristics, while the properties of material are influenced by the effect of time in service.

The factors that dictate the requirement in selecting structural materials are (a) operating loads—static, cyclic, or impact, (b) operating temperature, and (c) operating environment. While optimum mechanical properties viz., strength, ductility, toughness, etc., are important to ensure the reliability of a material's performance, but for the designed life the same at the specific condition of operation also becomes essential.

Today a large number of materials are available that can be employed as structural materials. They can be classified as (a) all-purpose materials and (b) materials possessing certain specific properties required for a particular application. Steels, cast irons, copper and copper-based alloys, light metals and alloys, etc. are all-purpose materials. The second group of materials also known as special materials is heat resistant, wear resistant, and corrosion resistant alloys, high temperature materials, and materials with high elastic properties. Materials with special electric and magnetic properties also come under special materials.

Imperfections in materials are often the origin of a large number of component failures. Imperfections can be in the form of nonmetallic inclusions, porosities, voids, and cracks. Although these imperfections can be detected by sophisticated nondestructive testing methods, they get into the component due to negligence both during manufacture as well as inspection.

For engineering applications, the limit of nonmetallic inclusions of various types generally does not find a mention in material specifications, however the same finds a clear mention in case of materials for aerospace applications. This leniency is often utilized both by the manufacturers as well as by the inspectors to their advantage. Ultimately when the failure occurs due to the high content of these inclusions, the user is not in a position to claim the damages due to inadequacy in the original specification. Therefore, it is imperative that the limit of nonmetallic inclusions is clearly specified even in the case of materials for general engineering purpose. The case studies that follow bring out the ill effects of selecting an unsuitable material for a given application. The consequences of such wrong decisions in material selection have caused heavy losses to the users.



4.1 CASE STUDY: FAILURE OF HIP-JOINT PROSTHESIS

4.1.1 Background

A prosthesis used for hip joint replacement failed within 7 days after its implantation. No information was made available on the prosthesis except that failure of this type of joints in such a short period is very unusual.

4.1.2 Experimental Results

Visual Examination: The hip joint prosthesis had broken into two parts across a hole as shown in Fig. 4.1.1. The fracture surface was coarse and crystalline (Fig. 4.1.2). Two punch marks were noticed on either side of the hole, which were joined by a crack on the surface of the implant (Fig. 4.1.3).

Fractographic Examination: Examination of the cleaned fracture surface under SEM revealed intergranular fracture features with secondary cracks (Fig. 4.1.4).

Metallographic Examination: Longitudinal section adjacent to the fracture was prepared for metallography. In the unetched condition, the sample showed a large population of elongated sulfide and globular oxide inclusions (Fig. 4.1.5). On etching with aqua regia, the sample revealed equiaxed grains of austenite with annealing twins and intergranular corrosion (Fig. 4.1.6).

Sample a little away from failure was subjected to a test for evaluating whether it was sensitized. For this purpose, the sample was boiled with 65% nitric acid for 30 min [1]. Examination of the sample after polishing and etching revealed intergranular corrosion (Fig. 4.1.7).

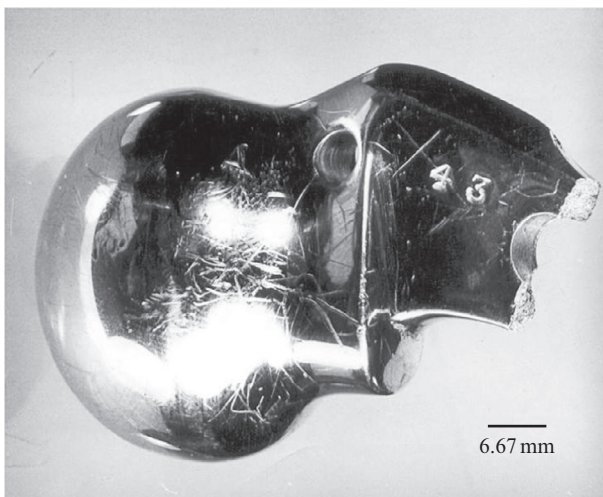


Fig. 4.1.1 Broken part of the hip joint prosthesis in the as-received condition.



Fig. 4.1.2 Fracture surface showing a coarse and crystalline fracture.

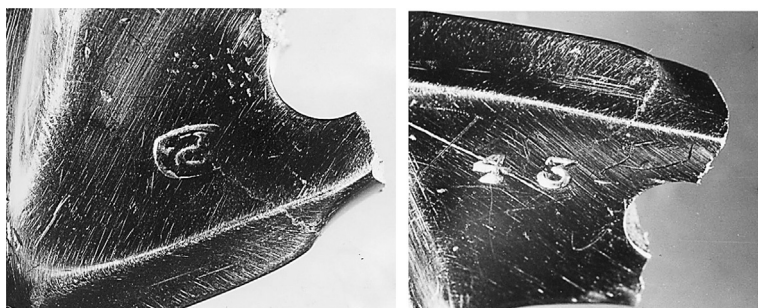


Fig. 4.1.3 Punch marks joined by a crack on the surface as seen on either side of the hole.

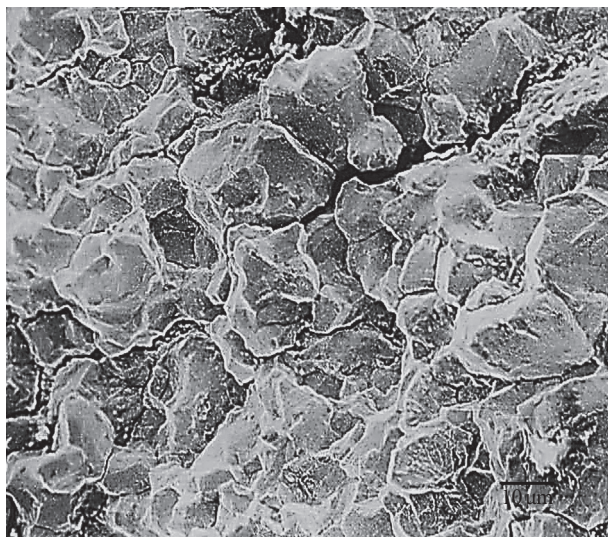


Fig. 4.1.4 SEM fractograph showing intergranular fracture with secondary cracks.

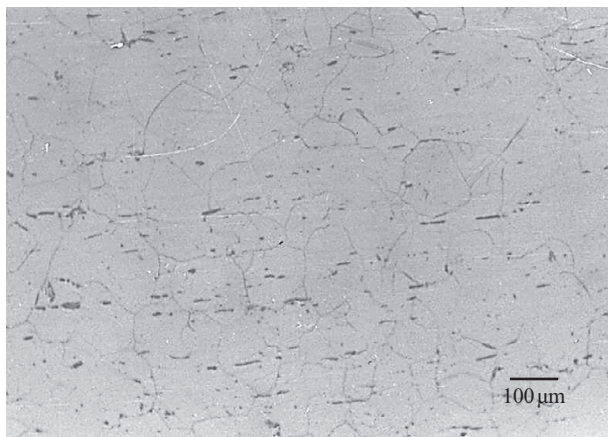


Fig. 4.1.5 Micrograph showing a large number of elongated sulfide and globular oxide inclusions. Unetched.

Hardness: Average Vickers hardness measured at 5 kg load near the failure was 283HV, while remote from failure was 228HV.

4.1.3 Chemical Analysis (Wt.%)

	C	Si	Mn	Ni	Cr	S	P	Fe
(i) Stem	0.16	0.40	1.73	7.5	15.50	0.054	0.07	Base
(ii) Head	0.12	0.55	1.20	8.03	19.00	0.01	0.06	Base

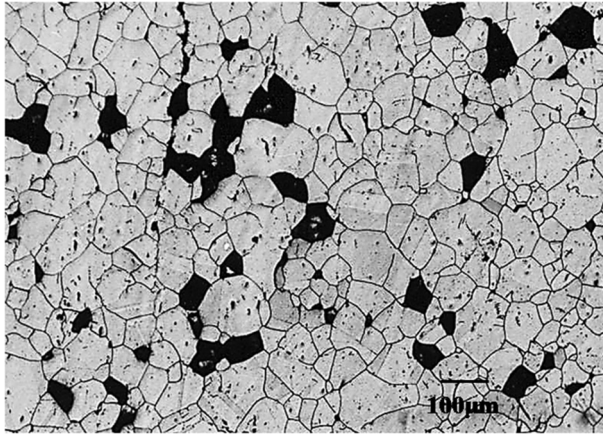


Fig. 4.1.6 Micrograph of longitudinal section adjacent to fracture revealing equiaxed grains of austenite with annealing twins and intergranular corrosion as evidenced by removal of polygonal grains in the microstructure. Etchant: Aqua regia.

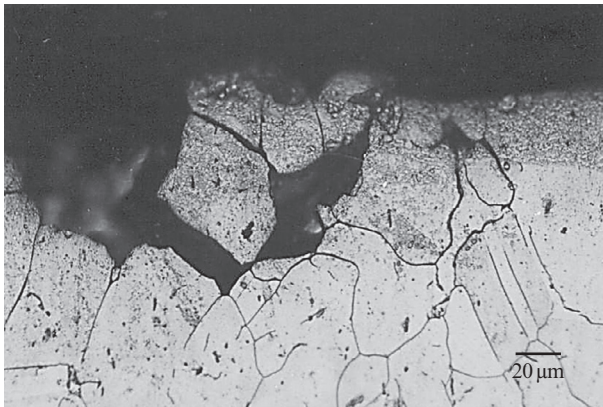


Fig. 4.1.7 Micrograph of a section of the sample that was subjected to test for sensitization showing intergranular corrosion. Etchant: Aqua regia.

4.1.4 Discussion

The hip joint prosthesis was made up of two parts having different chemical compositions, which were joined by welding. The spherical portion nearly conformed to austenitic stainless steel grade AISI 304, whereas the failed portion did not conform to any standard grades of stainless steel.

The failed part of the prosthesis was in sensitized condition as confirmed by boiling nitric acid test and so was susceptible to intergranular corrosion. The punch marks on either side of the hole have acted as sites for initiation of

corrosion when the prosthesis came in contact with biofluids after implantation. Nonmetallic inclusions, which were in large numbers in the component, also provided preferential sites for corrosion. The corroded implant broke inside the body when it was subjected to small stress due to movement of the hip portion. The premature failure of the sensitized and corroded prosthesis was bound to be in intergranular mode as evidenced by SEM fractography.

4.1.5 Conclusion

The hip joint prosthesis failed prematurely due to a combined action of

- (a) Improper selection of material,
- (b) Improper heat treatment leading to sensitized microstructure, and
- (c) Intergranular corrosion due to body fluids.

4.1.6 Remedial Measures

No metallic material is totally resistant to corrosion within living tissues. However, the following materials are more resistant to corrosion when in contact with body fluids and so are in common use for implants and prosthesis such as

- (i) 316L type austenitic stainless steel,
- (ii) Titanium and titanium alloys,
- (iii) Cast and wrought Co-Cr base alloys, and
- (iv) Tantalum where low resistance to deformation is required.

4.1.7 Reference

[1] ASTM Standard A262-98, practice C, Section 01, vol. 1.03.



4.2 CASE STUDY: FAILURE OF FIRING PIN OF 120 MM LOW RANGE MORTAR

4.2.1 Background

Firing pin assembly in the breach block of a 120 mm low-range mortar failed prematurely during user trials. The firing pin was made of maraging steel. The firing pin assembly was subjected to an axial compression stress of 780 MPa when fired at a rate of 5–6 rounds per minute, which is continued for 15 min at a time. During this exercise the temperature inside the mortar reaches to about 700–780°C. There is no means to measure the temperature of the firing pin assembly by the users.

The minimum mechanical properties required for the pin were as given below:

0.2% P.S: 1000 MPa, UTS: 1200 MPa, Elongation: 12%, Impact toughness (at ambient temperature): 30J.

The firing pin and other components of the assembly were also expected to be corrosion and oxidation resistant and have good fatigue strength.

4.2.2 Experimental Results

Visual Examination: The failed part of the firing pin in the as-received condition is shown in Fig. 4.2.1. The neck region was found to have been deformed.

Metallographic Examination: Longitudinal section of the sample showed a large number of globular oxide and elongated sulfide inclusions (Fig. 4.2.2).



Fig. 4.2.1 The failed firing pin in the as-received condition.

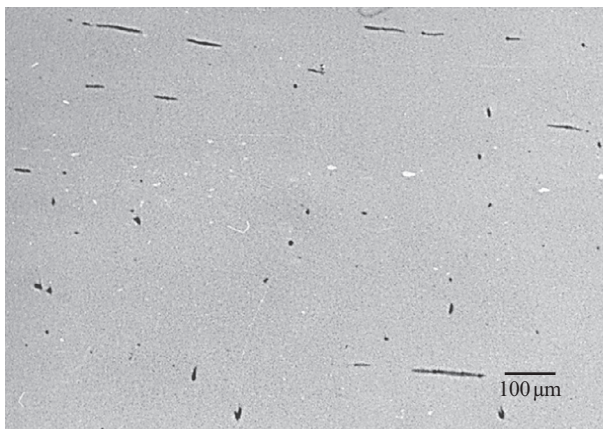


Fig. 4.2.2 Micrograph of the longitudinal section of the sample showing a large number of globular oxide and elongated sulfide inclusions. Unetched.

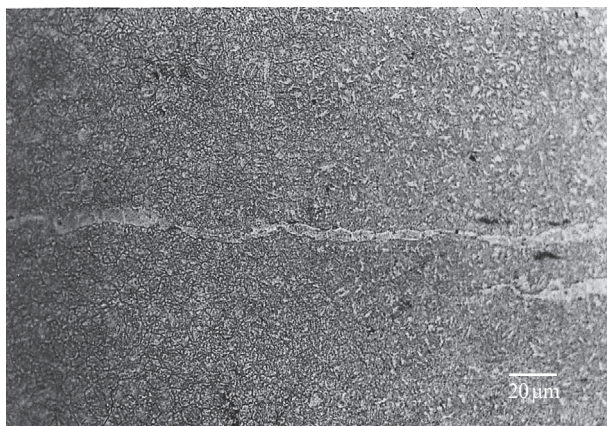


Fig. 4.2.3 Micrograph of the longitudinal section of the sample showing martensite and reverted austenite. Etchant: 2% nital.

On etching with 2% nital, the sample revealed martensite and reverted austenite (Fig. 4.2.3).

Hardness: Hardness of the sample varied widely from 500HV to 630HV.

4.2.3 Chemical Analysis (Wt.%)

C	Ni	Co	Mo	Ti	Al	Cr	Fe
0.01	17.41	9.70	4.48	0.79	0.09	0.10	Rem.

4.2.4 Discussion

The chemical composition of the firing pin conformed to maraging steel of grade nearly equivalent to 18Ni300. The microstructure consisting of martensite and reverted austenite could develop either during heat treatment or during service. The undesirable microstructure has caused differential hardness, which is detrimental to the load-bearing capability of the steel.

18Ni300 wrought maraging steel is solutionized at 820°C for 1 h for each 25 mm section thickness and air cooled, followed by aging at 480 °C for 3 h [1]. Maraging steel in peak aged and slightly over-aged condition has excellent combination of mechanical properties viz. strength and toughness. However, the steel is not recommended for use at temperature beyond 480°C mainly because the steel would experience overaging, resulting in deterioration in strength.

During firing the inner temperature of the mortar is said to be of the order of 700–750°C. This indicates that during sustained firing the firing pin assembly would experience relatively high temperature, which may exceed 500°C.

4.2.5 Conclusion

The firing pin was damaged due to deficiency in microstructure causing deterioration in strength, which in all probability had developed during service of the component at temperatures above the aging temperature of the maraging steel.

4.2.6 Remedy

Maraging steel is not the right choice for the intended application. Steels like 16Cr-2Ni martensitic stainless (SAE 431) steel that has a good combination of strength, toughness, corrosion, and oxidation resistance up to 650°C may be a better choice for firing pin application.

4.2.7 Reference

[1] M. Srinivas, A. Venugopal Reddy, Aero Steels: Part 2—High Alloy Steels, in: N. Eswara Prasad, R.J.H. Wanhill (Eds.), Aerospace Materials and Material Technologies, vol. 1, Indian Institute of Metals and Springer Nature, Singapore, 2017, pp. 173–198.



4.3 CASE STUDY: FAILURE OF SHAFT OF A GEARBOX

4.3.1 Background

The input shaft of a gearbox used in a winch failed prematurely into three pieces during operation. The shaft was stated to have been forged from a low alloy steel, machined, heat treated, and nitrided to achieve a case depth of 0.3–0.4 mm with a case hardness of 725HV. Due to the failure of the shaft, the bearings mounted on it were completely damaged.

4.3.2 Experimental Results

Visual Examination: The failed shaft in the as-received condition is shown in Fig. 4.3.1. Fracture at position ‘A’ was found to be completely rubbed, while fracture at position ‘B’ was relatively intact. The fracture at position ‘B’ showed a cup and cone type of fracture, whereas the fracture at position ‘A’ showed negligible deformation.

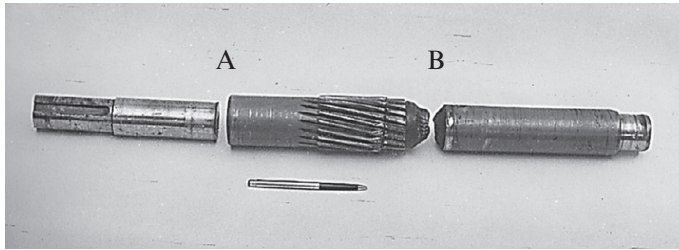


Fig. 4.3.1 Failed shaft in the as-received condition.

Fractographic Examination: On cleaning and examination of the fracture surface from position 'B' under SEM, the sample revealed predominantly dimpled rupture (Fig. 4.3.2). Fractography of the fracture surface from position 'A' could not be carried out due to severe rub marks and surface damage.

Metallographic Examination: Samples prepared for metallography showed a number of elongated sulfide and globular oxide inclusions (Fig. 4.3.3). On etching with 2% nital, the sample revealed a nitrided layer of case depth about 0.4 mm (Fig. 4.3.4). The core showed a tempered martensite microstructure with proeutectoid ferrite at prior austenite grain boundaries (Fig. 4.3.5).

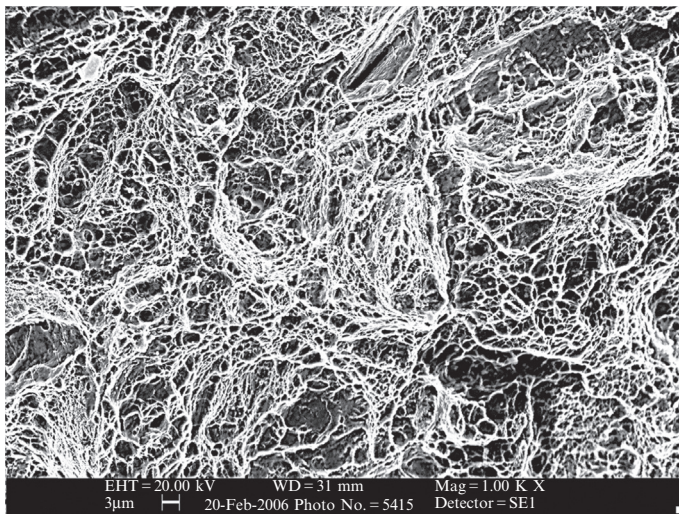


Fig. 4.3.2 SEM fractograph from the fracture surface at position 'B' revealed predominantly dimpled rupture.

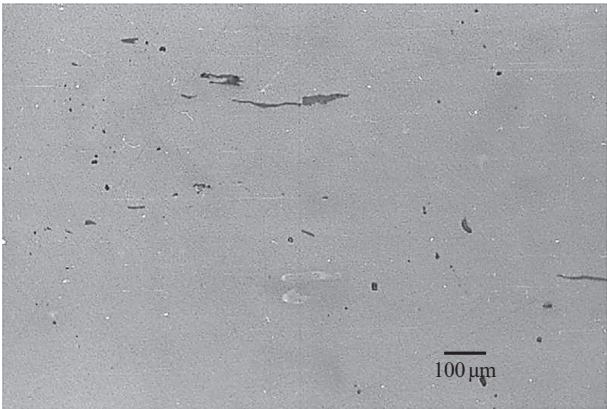


Fig. 4.3.3 Micrograph of the sample showing a number of elongated sulfide and globular oxide inclusions. Unetched.

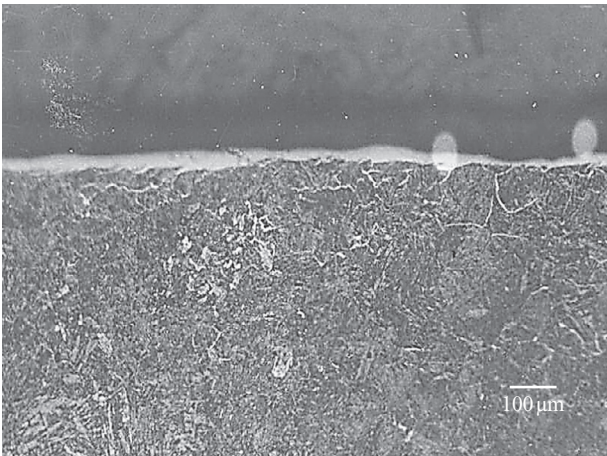


Fig. 4.3.4 Micrograph of the sample showing nitrided layer with a white etching layer of iron nitride on the surface. Etchant: 2% nital.

Hardness: The average Vickers micro hardness of the nitrided case was 650HV at 100g load and the average Vickers hardness of the core was 312HV at 10kg load.

4.3.3 Chemical Analysis (Wt.%)

C	Si	Mn	Cr	Ni	Mo	S	P	Fe
0.43	0.48	0.76	1.15	0.17	0.21	0.014	0.02	Balance

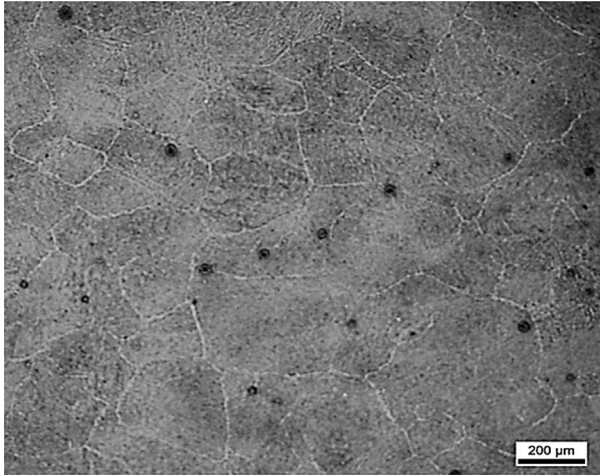


Fig. 4.3.5 Micrograph of the sample showing proeutectoid ferrite along prior austenite grain boundaries in a matrix of tempered martensite. Etchant: 2% nital.

4.3.4 Discussion

The chemical composition of the shaft nearly conformed to grade En19C of BS 970-1970. It is surprising to note that grade En19C, which is not recommended for nitriding, was used in this case. In the absence of other nitride-forming elements it is difficult to predict the consequences of chromium depletion in the steel due to nitride formation. Moreover, in such steels there is a high chance that iron nitride forms along with chromium nitride and render the case extremely brittle.

Due to sharp change in section size and also negligible deformation observed at position 'A', it is suspected that the primary failure had occurred here. The complete rubbing noticed on the fracture surface at position 'A' was indicative of the fact that the gear assembly was functioning even after the initial failure of the shaft at position 'A'. The failure at position 'B' was due to overload, which was a consequence of the initial failure.

4.3.5 Conclusion

The shaft had failed due to brittle nitride layer formed on the steel, which was unsuitable for nitriding.

4.3.6 Remedial Measures

En 41A steel of BS 970 Part 2-1970 may be used as it is recommended for nitriding [1].

4.3.7 Reference

[1] B.S. 970 Part 2:1970, Grade En 41A, British Standard Specification.

4.4 CASE STUDY: FAILURE OF 7.62 MM CARTRIDGE CASES

4.4.1 Background

A large number of 7.62 mm cartridge cases failed during proof testing of the ammunition leading to the rejections of the entire batch of ammunitions. Preliminary investigation by the manufacturers could not isolate any stage of manufacture that was responsible for the premature failure. The cartridge cases were made using a 70–30 brass (BS 2871, Part CZ 126) and used in a deep drawn condition.

4.4.2 Experimental Results

Visual Examination: Cartridge cases, a failed one along with a good one in the as-received condition, are shown in Fig. 4.4.1. A crack of about 15 mm length at a distance of about 8 mm from the base can be seen in the failed cartridge case (Fig. 4.4.1). There was no significant thinning of the wall near the crack.

Fractographic Examination: The crack was cut open and the fracture surface was cleaned and examined under SEM. Up to the mid thickness, the fracture appeared like lamellar tear followed by ductile fracture (Fig. 4.4.2). Extensive tear-like features with their longitudinal axis parallel to the direction of

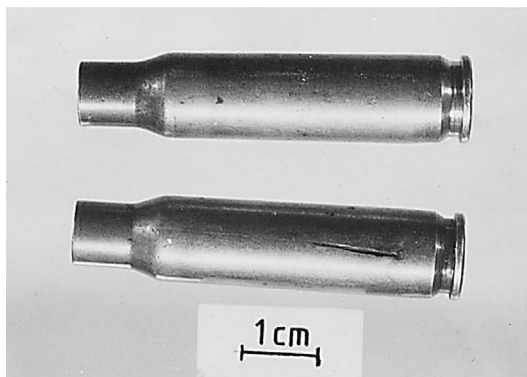


Fig. 4.4.1 Cartridge cases, a failed one along with a good one in the as-received condition.

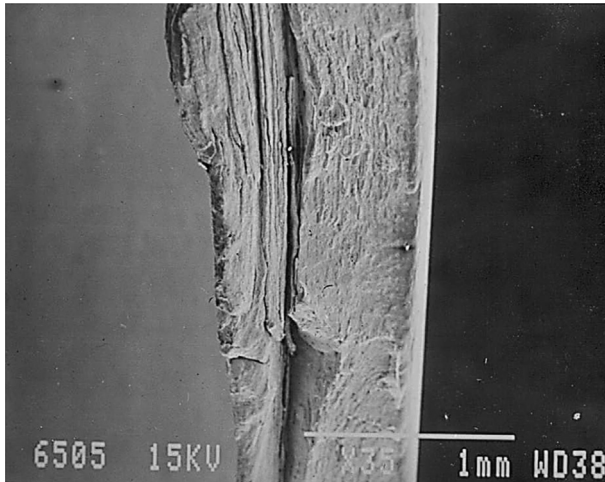


Fig. 4.4.2 SEM fractograph showing lamellar tear followed by ductile fracture.

drawing were observed near the origin (Fig. 4.4.3). The fracture surface showed predominantly a dimpled rupture (Fig. 4.4.4).

Metallographic Examination: Sections of both failed as well as good cartridge cases were prepared and examined under optical microscope. In the unetched condition, stringers of inclusions were seen both in failed as well as good cartridge samples (Fig. 4.4.5). The length of the stringers varied between 50 μm and 200 μm . On etching with acidified potassium



Fig. 4.4.3 SEM fractograph showing tear-like features with their longitudinal axis parallel to the direction of drawing near the origin.

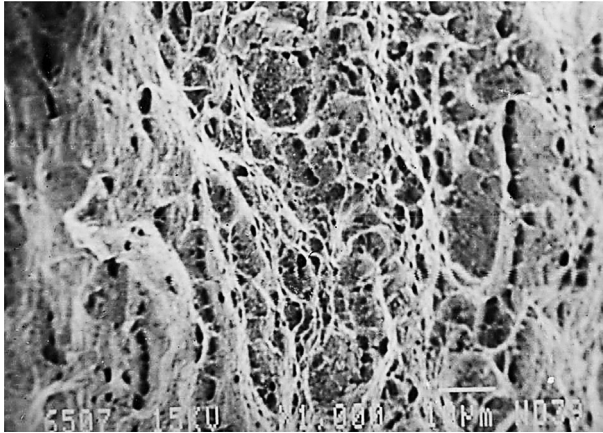


Fig. 4.4.4 SEM fractograph showing predominantly a dimpled rupture.

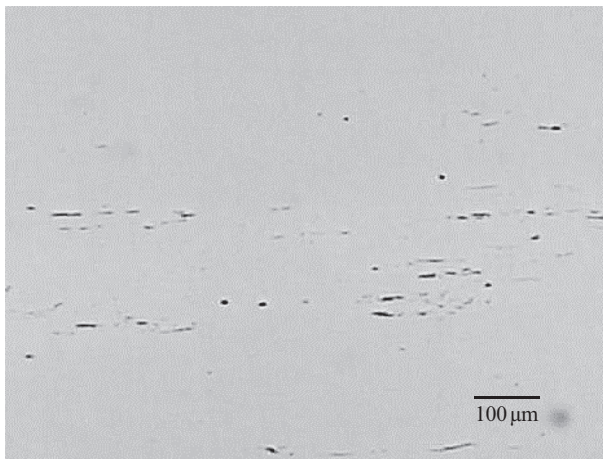


Fig. 4.4.5 Micrograph of longitudinal section of the sample showing stringers of inclusions. Unetched.

dichromate solution, the sample revealed deformed fine alpha grains, the extent of deformation increasing with increasing distance from the base. A typical microstructure at the mid-length of the cartridge case is shown in [Fig. 4.4.6](#). On analyzing the stringers by EDAX, the inclusions were found to contain sulfur, zinc, and copper ([Fig. 4.4.7](#)).

Hardness: Micro hardness scan from base to shoulder measured at 100 g and in an interval of 5 mm was as given in the following:

193HV, 182HV, 173HV, 185HV, 184HV, 157HV, 153HV, 142HV.



Fig. 4.4.6 Micrograph of longitudinal section of the sample showing deformed fine alpha grains. Etchant: Acidified potassium dichromate.

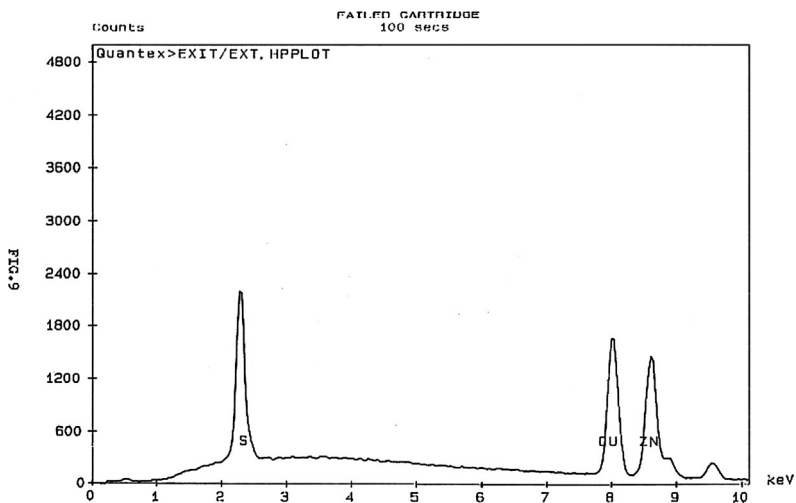


Fig. 4.4.7 EDAX of the inclusion showing presence of S, Cu, and Zn.

Chemical Analysis: The chemical analysis confirmed the cartridge alloy as 70–30 brass.

4.4.3 Discussion

The chemical composition conformed to the specified material (BS 2871, Part CZ 126). The microstructure showing varying degree of plastic deformation and uniform cross-sectional area at different diametrical sections from the base of the cartridge cases confirmed that there were no abnormalities in deep drawing and annealing operation. The presence of large stringers of inclusions and identical-sized tears observed on the fracture surface suggested that the stringers of inclusions caused the failure. EDS analysis indicated that the stringers are rich in sulfur and are probably zinc sulfide. The cracking of the cartridge cases can be attributed to the distribution of large-aligned stringers of sulfide inclusions and their encountering regions of maximum hoop stress.

4.4.4 Conclusion

The failure of the cartridge cases is due to the presence of aligned stringers of sulfide inclusions.



4.5 CASE STUDY: FAILURE OF FIN MOUNTING NUTS

4.5.1 Background

An aerospace weapon uses M8 screws and hexagonal lock nuts for fastening of the fins. Each fin requires four sets of screws and nuts. The torque used to fasten the screws was 20 Nm. In one such incident while application of torque two nuts got cracked. The nuts were made using a low alloy Ni–Cr–Mo steel of grade SAE 4340, hardened at 860°C, and tempered at 550°C. The nuts were finally cadmium plated and baked.

4.5.2 Experimental Results

Visual Examination: The failed nut in the as-received condition is shown in Fig. 4.5.1. The nut has fractured along the longitudinal axis.

Fractographic Examination: On examination under SEM, the fracture surface revealed dimpled rupture and separation along stringers of inclusions (Fig. 4.5.2).

Metallographic Examination: A section of the nut sample was prepared and examined under optical microscope. The sample showed a large number of

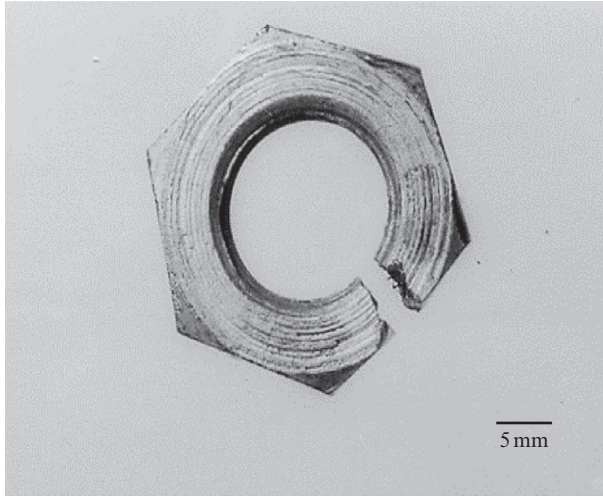


Fig. 4.5.1 Failed nut in the as-received condition.

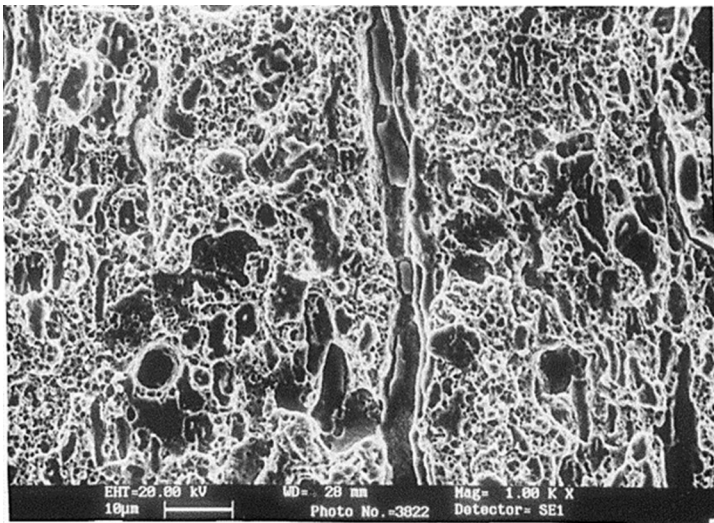


Fig. 4.5.2 SEM fractograph of the fracture surface showing dimpled rupture and separation along stringers of inclusions.

nonmetallic inclusions and also some large slag pockets (Figs. 4.5.3 and 4.5.4). On etching with 2% nital, the sample revealed a tempered martensite structure (Fig. 4.5.5).

Hardness: The average Vickers hardness measured at 20kg load was 379 HV.

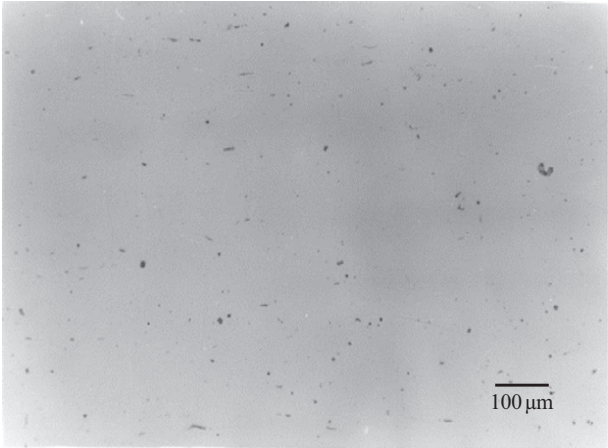


Fig. 4.5.3 Micrograph of the sample showing a large number of elongated sulfides and globular oxide inclusions. Unetched.

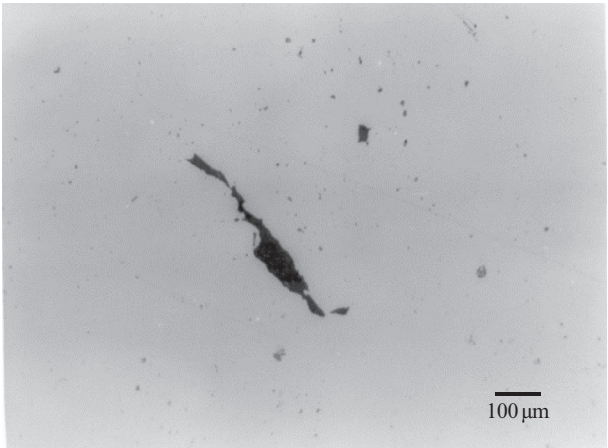


Fig. 4.5.4 Micrograph of the sample showing a large slag pocket. Unetched.

4.5.3 Chemical Analysis (Wt.%)

C	Si	Mn	Ni	Cr	Mo	S	P	Fe
0.40	0.42	0.71	1.60	1.20	0.23	0.03	0.025	Base

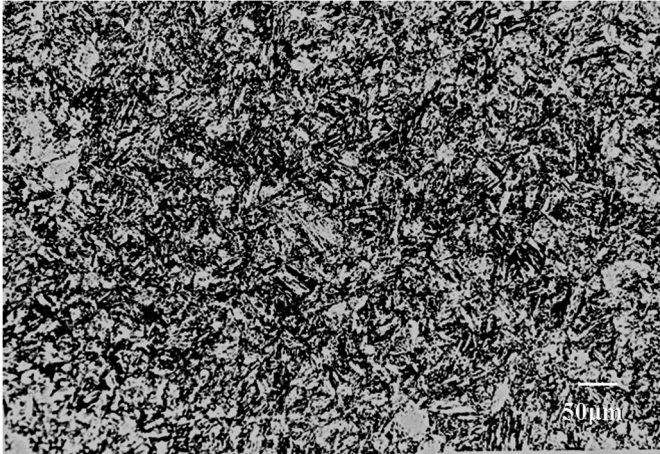


Fig. 4.5.5 Micrograph of the sample showing a tempered martensite structure. Etchant: 2% nital.

4.5.4 Discussion

Chemical analysis, microstructure, and hardness were nearly conforming to the specification. The inclusion contents of the nuts were abnormally high; however, the users have not specified any limit on the inclusion content in the steel. Large-sized inclusions of slag were other defects that affected the cleanliness of the material. The tearing of inclusions and dimpled rupture indicated that the failure was by overload initiated by inclusions. The high population of nonmetallic inclusions reduce the effective cross-sectional area of the product causing overload failure.

4.5.5 Conclusion

The nut had failed due to overload, which was caused by inadequate material cleanliness.



Manufacturing Defects

Surface discontinuities such as seams, laps, cold shuts, folds, etc. formed during metal working can later act as stress raisers of considerable severity during application of load in service. Subsurface or core discontinuities such as shrinkage cavities, porosities, blow holes, voids, and pipes formed during ingot solidification are known sources of weakness during subsequent fabrication or in service as these defects serve as sites for crack initiation. Imperfection resulting during metal working such as insufficient mechanical reduction while forging from ingot to bars can leave persistence of cast structure in wrought product. This would often go unnoticed in the manufacturing stage. If a part is machined out of such a bar, it will possess low ductility and toughness due to chemical and mechanical inhomogeneity. Chemical segregation carried down from ingot stage can result in banding of microstructure during hot working. Banding can cause discontinuities leading to cracking in service. Banding would also cause anisotropy of mechanical properties. Excessive segregation would lead to differential response to heat treatment causing nonuniform microstructure. For example, alternate layers of ferrite and martensite in a martensitic stainless steel would show vast difference in hardness of each layer and cause a weak interface that would be prone to cracking. Fabrication operations such as machining, grinding, welding, heat treatment, and electroplating can also introduce defects which will cause failures during service. Residual stresses produced by processes such as cold forming, machining, and grinding can leave a continuous tensile stress on the component resulting in stress-related failures.

Heat treatment is one of the important steps carried out in the manufacture of several engineering components. The schedule of heat treatment and the appropriate stage of component manufacture at which it has to be incorporated is based on sound metallurgical principles. Heat treatment carried out properly is expected to result in correct microstructure in order to get the most optimum properties. On the other hand, many of the component failures are the result of improper heat treatment imparted to the component.

Improper heat treatment can occur in a variety of forms, such as over-heating, inadequate tempering, low hardening temperature, faulty quenching practice, and unsuitable heat treatment method employed for a given alloy. Ignorance of the science and heat treatment of metals and alloys is another major cause of such failures.

The problems that arise due to improper heat treatment are: decarburization, distortion, tempered martensite embrittlement, temper embrittlement, quench cracking, low or high strength, poor toughness and ductility, corrosion, etc. Defective microstructure is produced due to improper heat treatment, which has a direct effect on the mechanical properties and corrosion resistance of metals and alloys. Heat treatment imparted to a component after surface coating is also significant from the point of view of the end use of the product.

A number of failures are caused due to poor planning, erroneous design, and faulty manufacturing and assembly. Although it is understood that sharp changes in cross sections and notches are centers of stress concentration, they still get into the design and manufacturing. Their presence can cause substantial increase in stress that is above the designed stress of any component resulting in failure. The failures can occur in such parts much before the assembly and service as well. Quench cracking during heat treatment of steel components is one such damage which initiates at sharp corners, cross sections, or notches left in a machined part. In a large number of cases, fatigue failures of cyclically loaded parts initiate at such unfavorable locations.

Welding has gained its importance as one of the principal methods of fabrication of metal structures and products. It is also economical in applications where repair of metallic products can enhance their lives.

Ferrous Metals: Mild steel containing up to 0.3% carbon has been the choice for fabrication of steel structures. However, the steel can show some brittleness in the heat-affected zone. Microalloyed steels containing 0.3% silicon, 1.0% manganese, $\leq 0.1\%$ carbon with small amounts of Nb, V, or Ti have found wide applications as structural materials due to their excellent weldability.

The carbon present in high carbon steels reacts with oxygen to form CO and CO₂, evolution of which can cause porosity in the weld. Deoxidants such as ferromanganese or ferrosilicon in the flux coating of the electrode can minimize this effect. Carbon picked up from the base metal into the weld zone causes formation of various metallurgical structures from pearlite through bainite to martensite depending upon cooling rates. Martensite formed in the heat-affected zone causes weld failures due to its high hardness and

brittleness. For welding alloy steels including stainless steels, choice of correct filler rod is essential. Formation of martensite and cracking are also problems associated with welding of alloy steels and ferritic high chromium steels. Preheating the base metal and post heat treatments is useful in preventing either the formation of martensite or in tempering the martensite already formed.

Austenitic stainless steel of 18/8 type is known to suffer from a defect termed 'weld decay'. The problem is termed as sensitization of stainless steels in processes other than welding and is well explained under Chapter 9 on 'Corrosion' in this book. During a fusion welding process, the area near the weld is likely to experience the vulnerable temperatures between 450°C and 850°C, thereby causing grain boundary carbide precipitation. Post weld solution treatment at 1050°C in order to redissolve chromium carbide followed by quenching can help in removing 'weld decay' of the part. Alternatively, addition of about 1.0% of Ti or Nb can tie up carbon of stainless steel due to their high affinity for it, making it unavailable for chromium to form chromium carbide along the grain boundary, thus avoiding 'weld decay'.

Cracking in the weld metal has been a problem encountered in alloy steels due to various reasons. Cracking at high temperature during solidification is called 'hot cracking'. Wrong choice of welding electrodes, large gaps between parts to be welded, pick up of embrittling elements viz. carbon, nitrogen, or hydrogen, rapid cooling rates, porosity, slag, or oxide inclusions, etc. are some of the reasons of weld cracking. Another type of cracking termed 'under-bead cracking' traces the weld bead contour. Hydrogen entrapped in the base metal or weld metal often cause under-bead cracking. Hydrogen pick up in the base metal or electrode can be from moisture in the environment, oil, grease, paint, or water acting as contaminants. The electrode also must be heated in an oven at 190°C for at least 24 h before using, which flushes out any hydrogen from it. Additionally, radial cracks, toe cracks, and internal cracks in the weld are causes of concern. Welding being a process of localized heating and cooling, expansion and contraction can cause shrinkage and distortion of the product. Cracking, shrinkage, and distortion should be prevented by taking suitable measures during and post welding. A notch or undercut should not be left at the edge of the weld as that could cause stress concentration.

Nonferrous Metals: Copper and copper base alloys although are weldable, the welder must be informed of certain facts about them to avoid failures. For example, copper has high affinity for oxygen and therefore deoxidized copper must be welded rapidly using gas metal arc process in order to avoid formation of oxides. Due to its high coefficient of thermal

expansion, copper tends to promote cracking in the weld during cooling especially if there is a restraint in the base metal. Copper base alloys that contain lead or zinc are difficult to weld since these elements volatilize at the welding temperature resulting in loss of strength. Mechanical work can improve the weld due to grain refinement.

Nonheat treatable aluminum alloys can be welded easily whereas heat treatable aluminum alloys are difficult to weld. Precipitation hardened aluminum alloy suffers loss of strength in the heat-affected zone due to precipitation of copper aluminide at the grain boundaries. Suitable post weld heat treatment can restore their strength. As aluminum alloys possess low strength at solidification temperatures, high rate of cooling induces strain in the weld metal causing hot cracking usually seen in the center of weld. Preheating of the base metal and choice of proper filler metal can minimize the effect of hot cracking in aluminum alloys.

Nickel base alloys are readily welded using similar filler rods as the base metal. Annealing after welding imparts the weld and the heat-affected zone uniform hardness. Filler rods must contain aluminum, titanium, or niobium in order to combine with oxygen or nitrogen dissolved in weld pool, thereby avoiding porosity. High-temperature sulfide embrittlement has been found in nickel alloys due to surface contamination of base metal with sulfur during welding.

The most important aspect of titanium alloys is to avoid the formation of oxides. The same is true with all other reactive metals viz., zirconium, magnesium, and tantalum. In order to prevent embrittlement, titanium alloys must be free from surface contamination and welding has to be carried under inert gas atmosphere.

A weld has to be examined visually and by subjecting it to a nondestructive test. Sometimes a destructive test and field test may also be necessary in order to approve the soundness of weld. Repair weld of large size components in the field of operation or maintenance bay makes it difficult to carry out a post weld inspection except by visual examination. This becomes a cause for weld failure of repair welded parts.



5.1 CASE STUDY: FAILURE OF TRACK PINS OF TRACK LINK

5.1.1 Background

A number of track pins of track link in a prototype main battle tank were reported to have failed during the trial run of the vehicle. The user's reported

that 19 pins out of 170 double pin track assembly broke within 550 km run though the expected life was about 3000 km. The pins were required to be manufactured from En 24 grade steel of BS 970—1955 and heat treated to a hardness level of 310–340 HB. The pins were manufactured and supplied by some trade firms.

5.1.2 Experimental Results

Visual Examination: The failed track pin sample in the as-received condition is shown in Fig. 5.1.1. The pin was found to have broken at the edge of a rectangular slot on the surface at one end. Broken parts of the rubber bush were seen sticking to surface of the pin. Fracture surface showed chevron marks converging at the central portion of the edge of the slot (Fig. 5.1.2).

Fractographic Examination: On examining the cleaned fracture surface under SEM it showed quasi-cleavage fracture features with secondary cracks (Fig. 5.1.3).

Metallographic Examination: Longitudinal and transverse sections of the pin sample were prepared for optical microscopic examination. In the un-etched condition, the sample showed a number of elongated sulfide and globular oxide inclusions. On etching with 2% nital, the transverse section revealed a typical cast structure with fine dendrites (Fig. 5.1.4). The longitudinal section showed a banded structure. At higher magnification the bright band showed spheroidized carbides and ferrite while the dark band revealed tempered martensite (Fig. 5.1.5).

Hardness: Average Vickers hardness of the failed sample at 30 kg load was 400 HV.

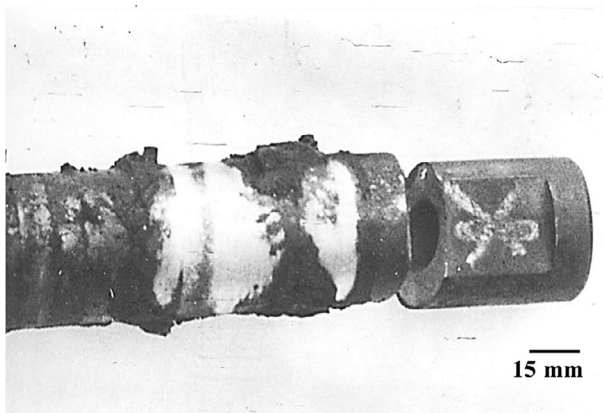


Fig. 5.1.1 The failed track pin sample in the as-received condition.

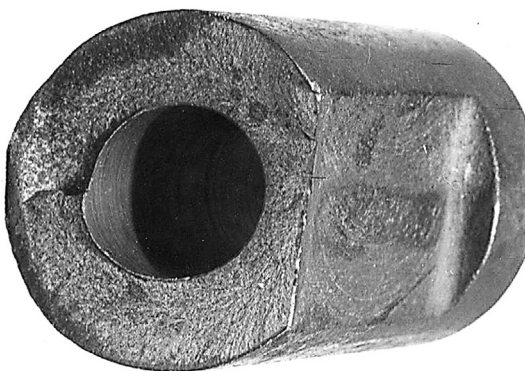


Fig. 5.1.2 Fracture surface showing chevron marks converging at the central portion of the edge of the slot

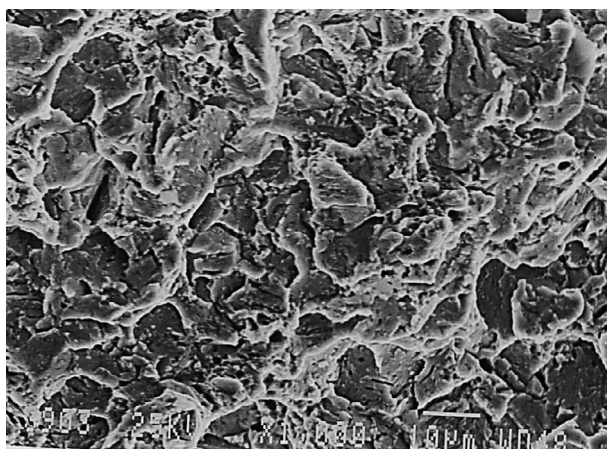


Fig. 5.1.3 SEM fractograph showing quasi-cleavage fracture features with secondary cracks

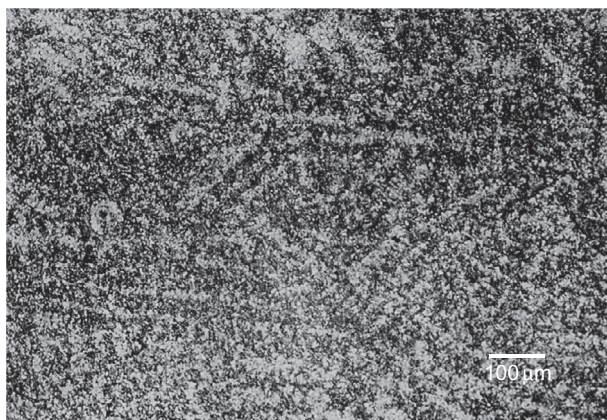


Fig. 5.1.4 Micrograph of the transverse section revealing a typical cast structure with fine dendrites. Etchant: 2% nital.

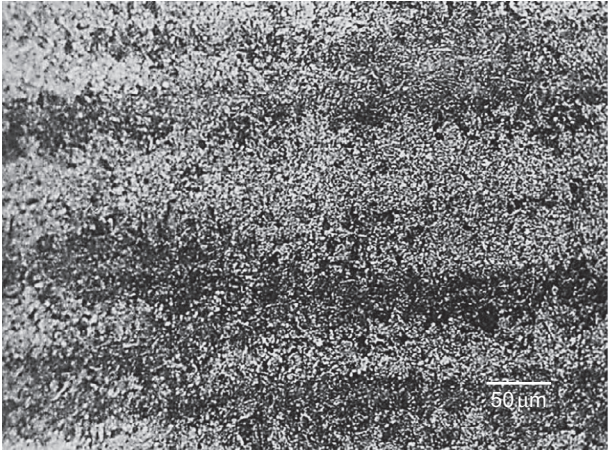


Fig. 5.1.5 Micrograph of the longitudinal section showing spheroidized carbides and ferrite in the bright band and tempered martensite in the dark band. Etchant: 2% nital.

5.1.3 Chemical Analysis (Wt.%)

C	Si	Mn	Ni	Cr	S	P	Fe
0.39	0.62	0.76	1.33	0.91	0.01	0.02	balance

5.1.4 Discussion

The chemical composition suggested that the failed track pins were nearly conforming to En24 grade steel. The quasi-cleavage fracture was indicative of the fact that the failure had occurred at relatively high strain rate, which is possible under the influence of a stress raiser. Here insufficient fillet radius at the edge of the slot could act as the requisite stress raiser. The persistence of cast structure indicated insufficient reduction from stages of cast ingot to the rolled pin. Chemical inhomogeneity due to the cast structure was responsible for banded microstructure varying from carbides and ferrite to tempered martensite after the heat treatment of the pins. Higher hardness of the pins as compared to the specification again indicated improper heat treatment or the initial microstructure not responding to the specified heat treatment.

The presence of cast structure, chemical inhomogeneity, and high hardness resulted in inadequate ductility. The radius provided at the edge of the slot would probably be insufficient for the altered condition of the steel.

5.1.5 Conclusion

The failure of the track pins was due to deficiencies in processing of the steel.

5.1.6 Recommendations

- (i) Use wrought (sufficiently reduced) En24 grade steel bar for making the track pins.
- (ii) Heat treat to attain the specified mechanical properties. The pins may be released for assembly only after checking the mechanical properties by random sampling.
- (iii) Consider the need to increase the radius at the edge of the slot if that does not affect the design seriously.



5.2 CASE STUDY: FAILURE OF TRACK SHOE OF AN INFANTRY COMBAT VEHICLE

5.2.1 Background

Track shoe of an infantry combat vehicle broke into pieces during acceptance running trials. The track shoe was forged from a steel billet of grade 20 XГCHM to specification Ty 14-1-2320-78 (Russian), and was hardened and tempered to a hardness level 450-475 BHN.

5.2.2 Experimental Results

Visual Examination: A part of the failed track shoe is shown in Fig. 5.2.1. The fracture surface was found to be dark along the length to a depth of 2mm

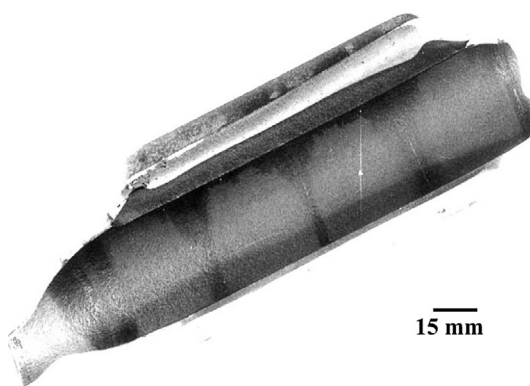


Fig. 5.2.1 Part of the failed track shoe in the as-received condition.

near the inner surface (Fig. 5.2.2). The rest of the fracture surface showed chevron marking.

Fractographic Examination: On examining the cleaned fracture surface under SEM it showed fracture steps initiating from the inner surface (Fig. 5.2.3). The dark region showed adherent oxide debris which was masking the fracture features (Fig. 5.2.4). On further cleaning it with oxide remover and examining, the fracture surface revealed intergranular fracture with secondary intergranular cracks (Fig. 5.2.5). The remaining portion of the fracture surface revealed quasi-cleavage fracture (Fig. 5.2.6).

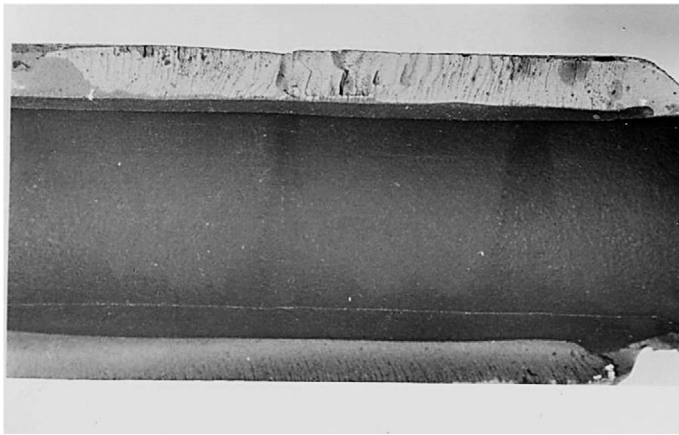


Fig. 5.2.2 Fracture surface of the failed component showing dark color to a depth of 2mm near the inner surface.



Fig. 5.2.3 SEM fractograph showing fracture initiating from the inner surface.

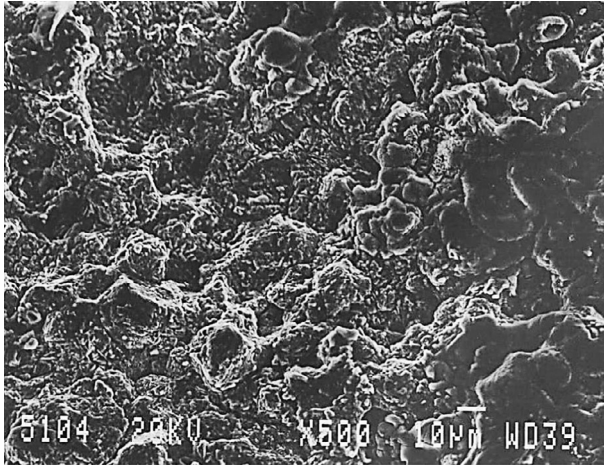


Fig. 5.2.4 SEM fractograph showing adherent oxide layer on the dark region of the fracture surface.

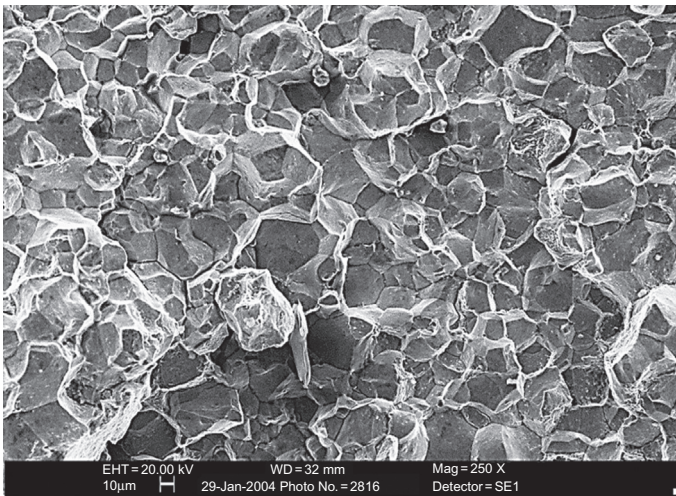


Fig. 5.2.5 SEM fractograph showing intergranular fracture with secondary intergranular cracks.

Metallographic Examination: Longitudinal and transverse sections of the sample were prepared for optical microscopic examination. In the un-etched condition, the sample showed a number of elongated sulfide and globular oxide inclusions (Fig. 5.2.7). On etching with 2% nital, the transverse section revealed decarburized layer near the dark region of the fracture

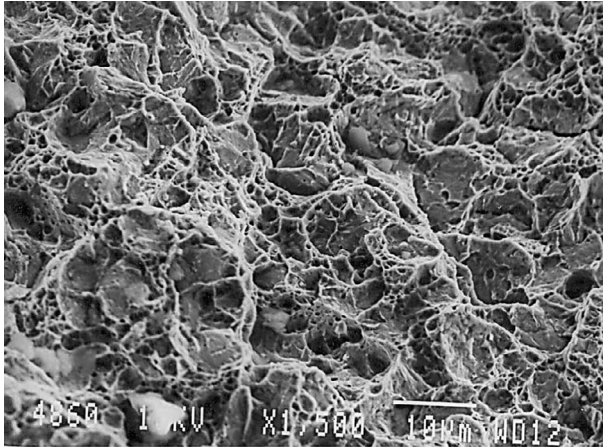


Fig. 5.2.6 SEM fractograph showing quasi-cleavage fracture on the remaining fracture surface.

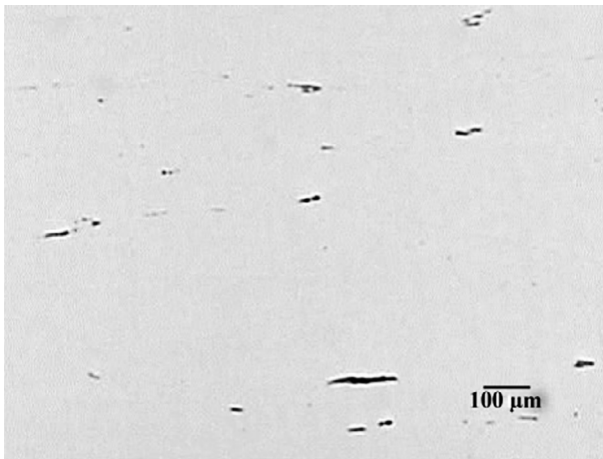


Fig. 5.2.7 Micrograph of the longitudinal section of the sample showing elongated sulfide and globular oxide inclusions. Unetched.

surface and intergranular cracks (Fig. 5.2.8). The longitudinal section showed a tempered martensite structure (Fig. 5.2.9).

Hardness: Microhardness scan from the inner surface toward the outer surface showed hardness variation from 300 HV to 500 HV. Average Vickers hardness of the track shoe away from the decarburized layer was 588 HV.

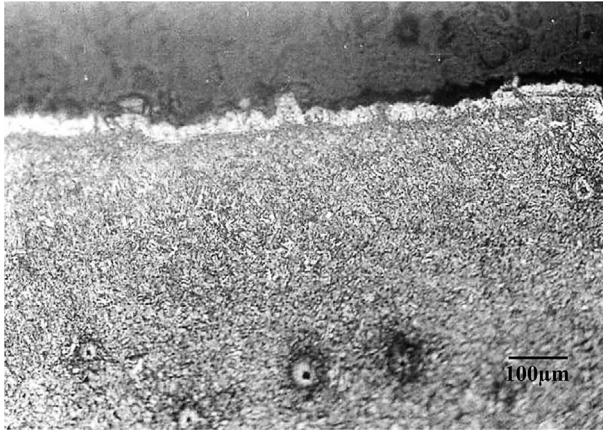


Fig. 5.2.8 Micrograph of transverse section of the sample showing decarburized layer and intergranular cracks near the surface. Etchant: 2% nital.

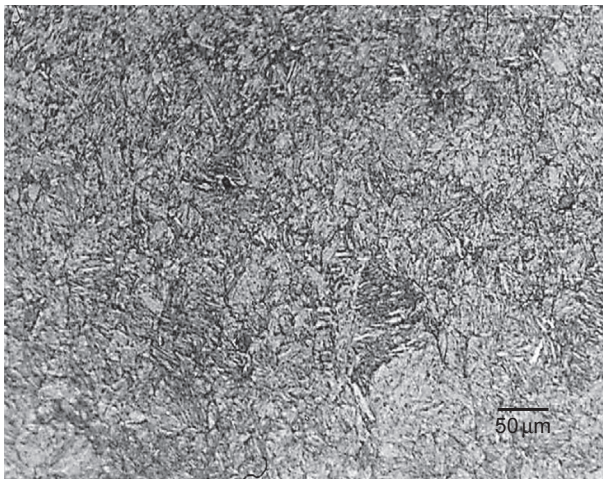


Fig. 5.2.9 Micrograph of the longitudinal section of the sample showing a tempered martensite structure. Etchant: 2% nital.

5.2.3 Chemical Analysis (Wt.%)

Failed Track Shoe

C	Si	Mn	Ni	Cr	Mo	S	P	Fe
0.35	1.20	0.38	0.24	1.40	0.17	0.02	0.014	balance

Russian Grade 20ХГЧМ (Specified Composition)

C	Si	Mn	Ni	Cr	Mo	S	P	Fe
0.17– 0.23	0.90– 1.20	0.80– 1.10	0.80– 1.10	0.80– 1.10	0.20– 0.30	0.025 max.	0.025 max.	Bal.

5.2.4 Discussion

Chemical analysis indicated that the material of the failed track shoe did not conform to the specified grade. The heat treatment imparted to the changed steel had resulted in an increase in hardness of the component compared to the stipulated range, and a consequent decrease in the ductility of the steel. The hardening and tempering parameters used for the composition of steel different from the specified one may not have been suitable to generate the required mechanical properties.

The dark adherent oxide on the fracture surface near the inner surface masking intergranular fracture features indicated that this was a preexisting crack in the component that was inherited from a stage prior to heat treatment. Oxidation and consequent decarburization of the regions adjacent to the fracture surface suggest that the preexisting intergranular crack was exposed to the furnace atmosphere during austenitization. The evidences thus point out that the most probable stage of development of the crack in the track shoe was during material processing.

The final failure of the component has occurred due to high strain rate under the stress-raising influence of the preexisting crack in a brittle mode. The higher hardness and lower ductility of the changed steel has aided in the brittle failure of the component by lowering its brittle fracture energy.

5.2.5 Conclusion

The failure of the track shoe was because of combined effects of incidence of crack during processing and nonconformance of chemical composition affecting the heat treatment and properties.



**5.3 CASE STUDY: FAILURE OF CONTROL LEVER
OF FUEL-METERING UNIT OF AN AEROENGINE**

5.3.1 Background

Control lever of fuel-metering unit of an aeroengine broke in service. The control lever is attached with control rod connecting throttle in the cockpit.

The component experiences a maximum working temperature of 60–80°C. It is stated to have completed only 534:50 hours of life since new while the total technical life was 4800 h.

5.3.2 Experimental Results

Visual Examination: The failed component in the as-received condition is shown in Fig. 5.3.1. The fracture surface showed two distinct planes. One fracture surface showed dark layer of grease and dust smeared on it and other was brighter and relatively smooth.

Fractographic Examination: The fracture surface was cleaned with dilute oxide remover followed by cleaning in acetone ultrasonically and examined under SEM. The fracture surface showed three distinct regions. The originally dark region of the fracture surface appeared granular, brighter region appeared smooth and the shear lip region appeared fibrous (Fig. 5.3.2A and B). At higher magnification, the granular part of the fracture surface revealed a predominantly transgranular cleavage fracture (Fig. 5.3.3), the smooth fracture surface revealed a quasi-cleavage fracture (Fig. 5.3.4) and the final failure region of about 500 μm width showed dimpled rupture (Figs. 5.3.5).

Metallographic Examination: Longitudinal section of the failed component taken from the elbow portion of the component was prepared for metallographic examination. In the unetched condition, the sample showed a few preexisting cracks filled with oxides near the surface (Fig. 5.3.6). On etching with aqua-regia decarburized layer was seen around one of the cracks (Fig. 5.3.7). The cracks were found to be propagating along the interface

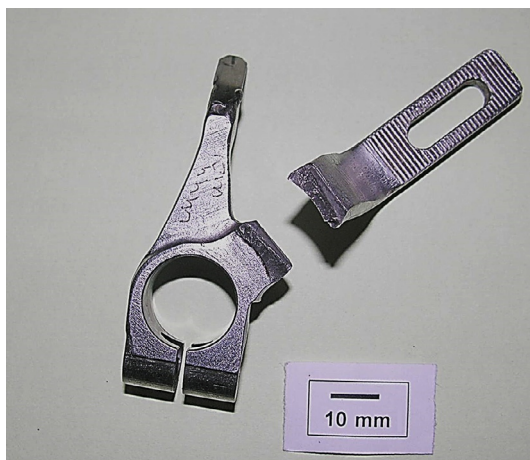


Fig. 5.3.1 The broken control lever in the as-received condition. The two distinct fracture surfaces with a prominent demarcation of change in plane are seen.

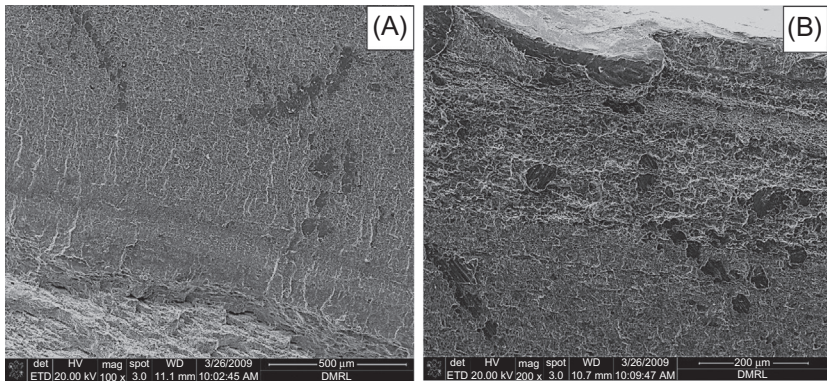


Fig. 5.3.2 (A) SEM fractograph of the fracture surface showing granular feature in the dark region and smooth fracture in the bright region. (B) SEM fractograph of the final failure region showing fibrous fracture.

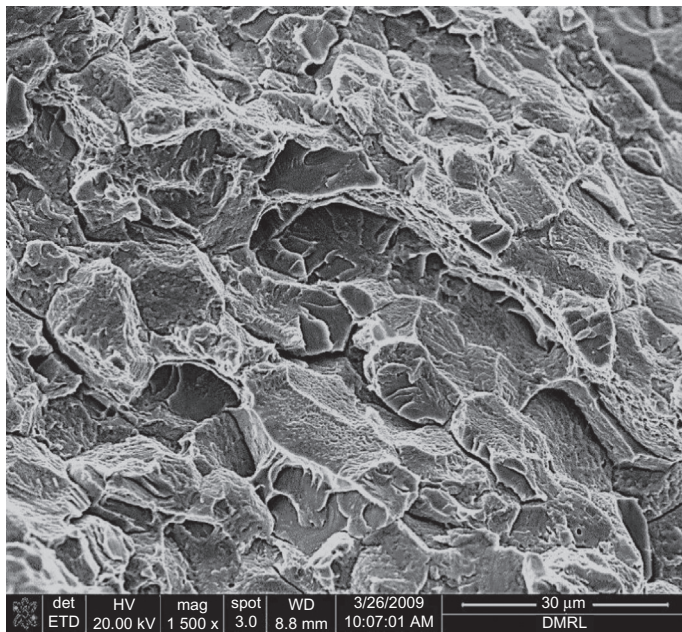


Fig. 5.3.3 SEM fractograph of the fracture surface with granular appearance showing predominantly transgranular cleavage fracture.

of martensite and δ -ferrite, the two major phases observed in the microstructure (Fig. 5.3.8). The microstructure showed grains tracing the contour of the profile of the component. The microstructure revealed tempered martensite and δ -ferrite (Fig. 5.3.9). The volume fraction of δ -ferrite was nearly 15%–20%.

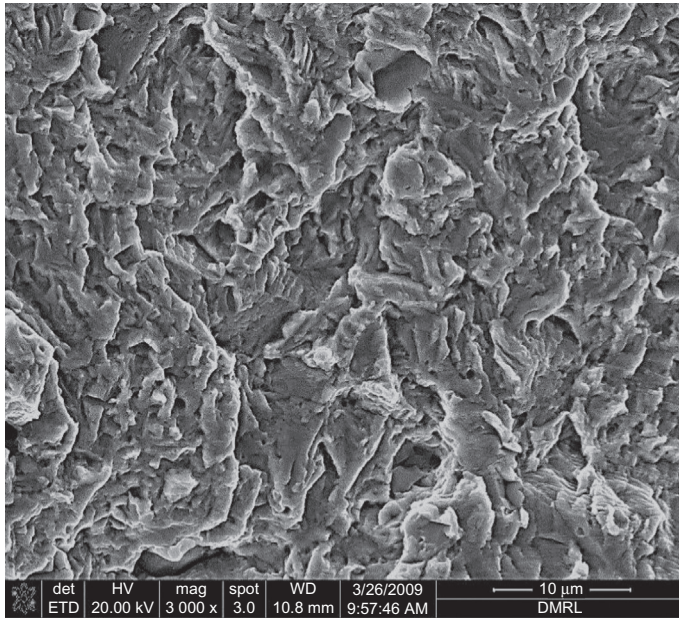


Fig. 5.3.4 SEM fractograph of the smooth fracture region showing quasi-cleavage fracture.

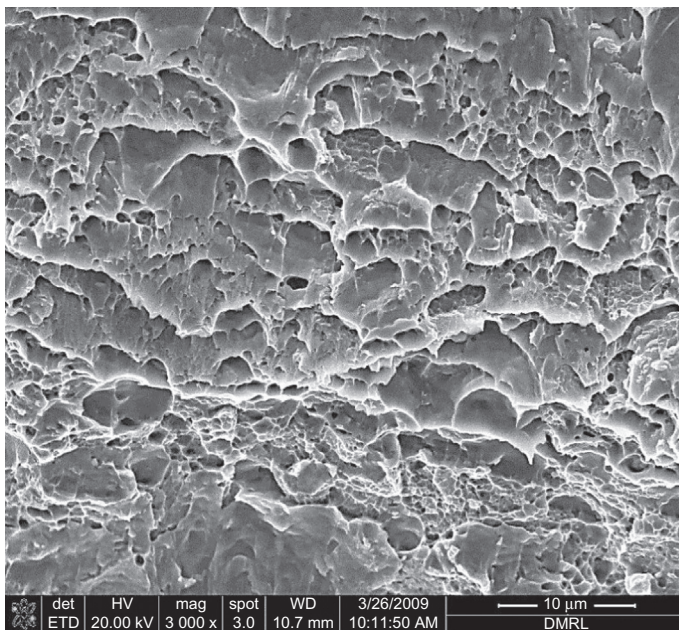


Fig. 5.3.5 SEM fractograph of the final fracture region showing dimpled rupture.

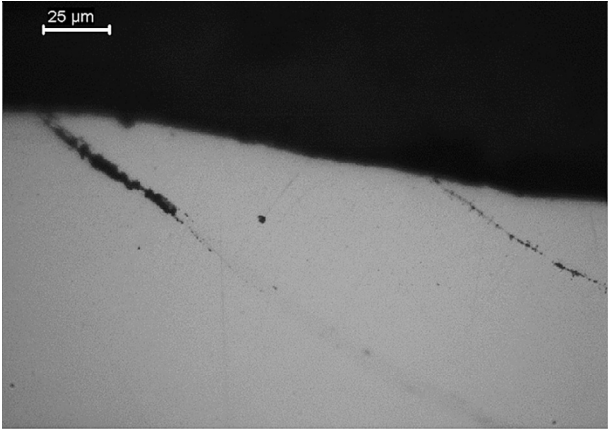


Fig. 5.3.6 Micrograph of a sample away from fracture showing preexisting Oxide-filled cracks near the surface. Unetched.

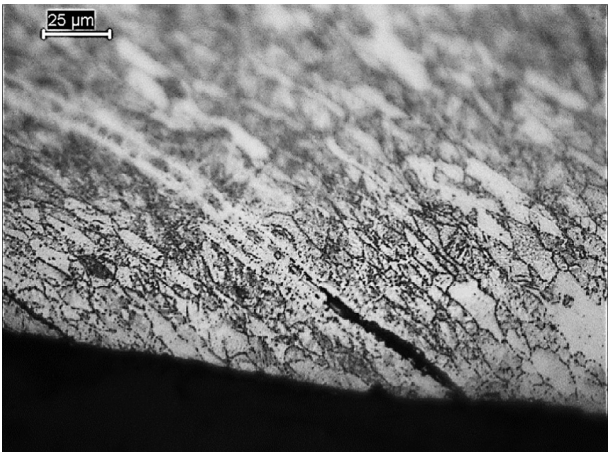


Fig. 5.3.7 Micrograph of the sample showing decarburization around the preexisting wide crack. Etchant: aqua regia.

Hardness: Average Vickers micro hardness and bulk hardness of the component was measured on the sample used for metallography and the results obtained are given here:

δ-ferrite	220 HV/50 g
tempered martensite	330 HV/50 g
Bulk hardness	315HV/10kg

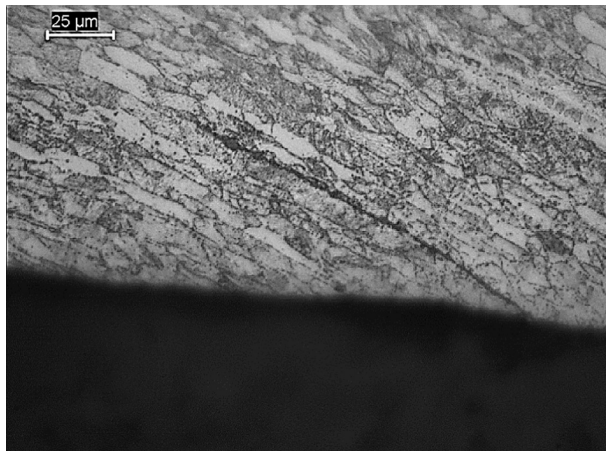


Fig. 5.3.8 Micrograph of the sample showing preexisting transgranular crack propagating along the interface of martensite and δ -ferrite. Etchant: aqua regia.



Fig. 5.3.9 Micrograph of a section of the sample showing aligned δ -ferrite in a matrix of tempered martensite. Etchant: aqua regia.

5.3.3 Chemical Analysis (Wt.%)

C	Mn	Si	Cr	Co	W	Fe
0.135	1.22	0.50	19.42	1.60	2.91	Base

5.3.4 Discussion

The chemical composition of the material does not conform to any specified grade, however it can be classified under martensitic stainless steel. From the modified Schaeffler diagram by Schneider, it can be observed that the composition of the material falls within the phase field of austenite + δ -ferrite + martensite [1]. Therefore, it can be predicted that when quenched from austenitization temperature to room temperature, the steel would form a microstructure comprising of martensite with considerable amount of retained austenite and δ -ferrite. Microstructure and hardness indicated that the component was used in a hardened and high temperature (about 650°C) tempered condition. The tempering temperature was high enough to transform retained austenite into ferrite and carbide. Hence, the microstructure that consisted of tempered martensite and δ -ferrite is on expected lines.

The microstructure revealed grain flow tracing the contours of the profile of the component, indicating that the component was a forged product. The presence of surface cracks with oxides suggested that they were formed during forging. The decarburized layer around the crack also suggested that the cracks were existing in the material prior to austenitization treatment. The presence of grease and dust on one part of the fracture surface indicated that it was a preexisting part of the crack that was exposed to the environment when the lever was operational.

The evidences revealed that at the region of failure there preexisted a forging crack. The stress-raising influence of the preexisting crack caused its propagation under operational load ultimately resulting in its failure. The high chromium and tungsten content in the δ -ferrite makes it a brittle phase in martensitic stainless steel [2]. The cleavage mode of failure is due to the presence of δ -ferrite in the microstructure and brittleness caused due to tri-axial state of stress at the preexisting crack.

5.3.5 Conclusion

The control lever of the fuel-metering unit had failed due to the presence of a preexisting crack that formed during forging. The propagation of the crack to final failure was aided by the microstructure of the material that contained a high volume fraction of δ -ferrite in a tempered martensite matrix.

5.3.6 References

- [1] K.P. Balan et al., Mater. Sci. Technol., 15 (7) (1999) 798–802.

- [2] F.B. Pickering, Stainless Steel, 84, The Institute of Metals, London, 1985, pp. 2–28.



5.4 CASE STUDY: FAILURE OF DRILL RODS FOR MINING

5.4.1 Background

One batch of a large number of tungsten carbide tipped drill rods used as mining tool was reported to be failing in service prematurely. The drill rods were manufactured using steel grade 1C-1Cr-0.25Mo and heat treated as specified by the manufacturer of the steel.

5.4.2 Experimental Results

Visual Examination: Two failed drill rods in the as-received condition are shown in Fig. 5.4.1, and an unused one from the same batch of manufacture is shown in Fig. 5.4.2. On visual examination it was observed that in all the samples received for failure analysis the collar-shank fillet was sharp and the surface at this junction was quite rough (Fig. 5.4.2). From Fig. 5.4.1 it can be seen that the failure had occurred near the collar-shank junction in all the samples. The fracture surfaces were found rubbed, damaged, and rusted in all the samples (Fig. 5.4.3).

Fractographic Examination: The fracture surfaces were cleaned with an oxide remover solution followed by ultrasonic cleaning in acetone. The cleaned fracture surfaces were examined under SEM. Fracture surfaces of all the three samples showed chevron marks diverging from a region near the collar-shank junction. A typical fracture origin is shown in Fig. 5.4.4. The region near the fracture origin showed some adherent oxides and pits (Fig. 5.4.5). The fracture surface revealed transgranular cleavage fracture. A typical fractograph is presented as Fig. 5.4.6.

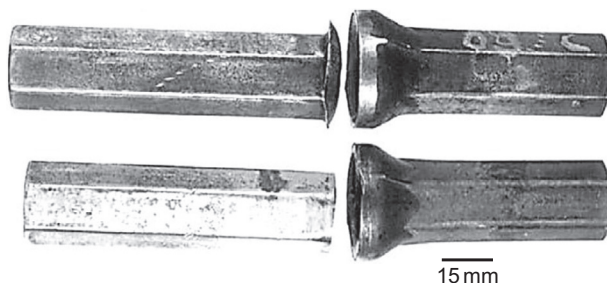


Fig. 5.4.1 Failed drill rods in the as-received condition.



Fig. 5.4.2 An unused drill rod from the same batch of manufacture in the as-received condition.

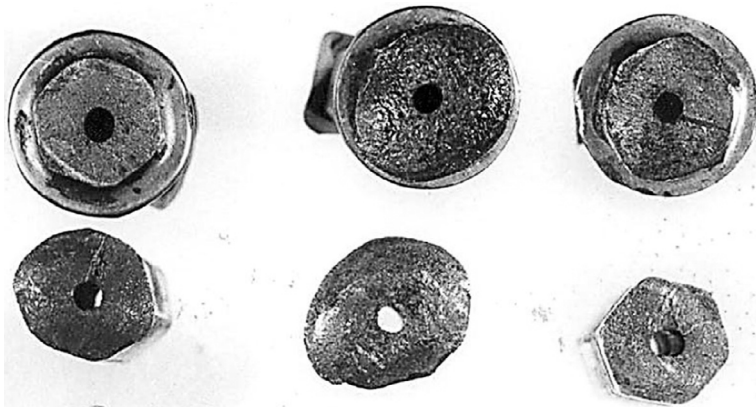


Fig. 5.4.3 Fracture surfaces showing rubbed, damaged, and rusted surfaces.

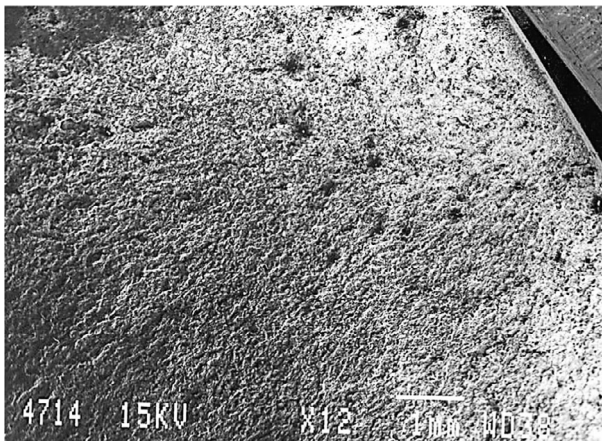


Fig. 5.4.4 SEM fractograph of the fracture surface showing a typical crack origin at the collar-shank junction.

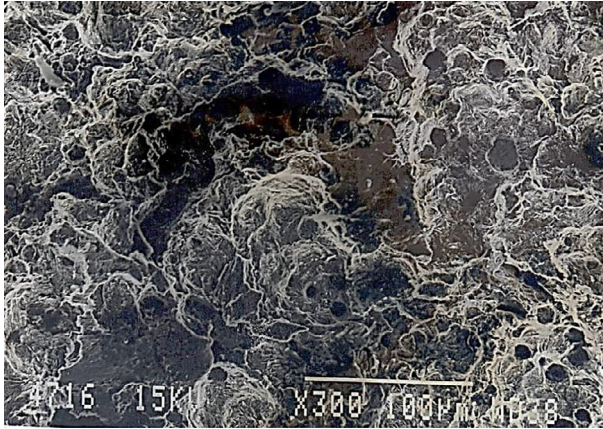


Fig. 5.4.5 SEM fractograph of the fracture surface near the crack origin showing some adherent oxides and pits.

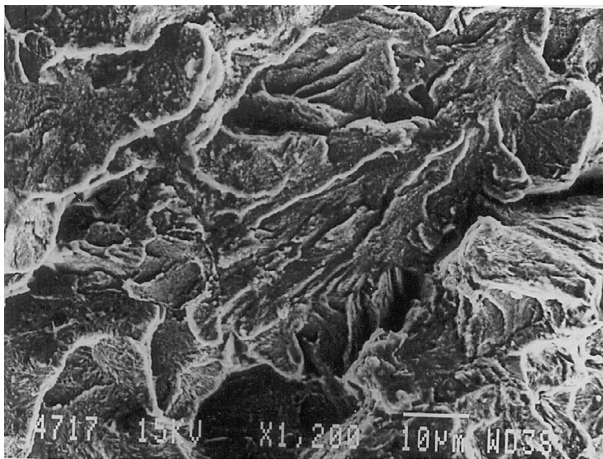


Fig. 5.4.6 SEM fractograph of the fracture surface showing transgranular cleavage fracture.

Metallographic Examination: Longitudinal sections from near the failed region were prepared from all the drill rod samples and examined under optical microscope. On etching with 2% nital, the samples revealed ferrite and lamellar pearlite structure. A representative microstructure is shown in [Fig. 5.4.7](#). The unused drill rod sample was also examined under optical microscope after preparing a longitudinal section of the sample from a similar area as that of the failed samples. The fillet contour at the collar-shank junction showed sharp nicks ([Fig. 5.4.8](#)). A crack was also observed to have initiated from one of the nicks and propagated in a transgranular mode

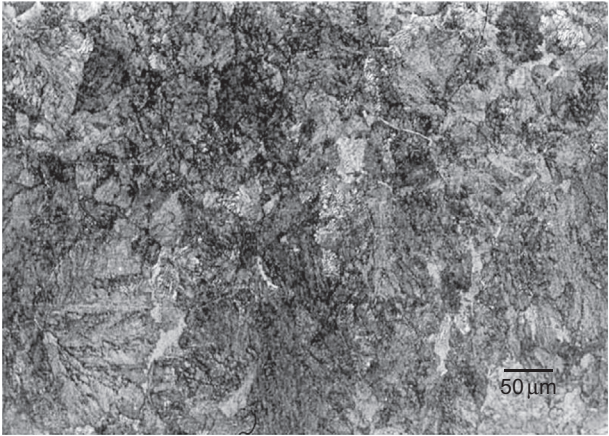


Fig. 5.4.7 Micrograph of longitudinal section of the failed drill rod sample showing ferrite and lamellar pearlite structure. Etchant: 2% nital.

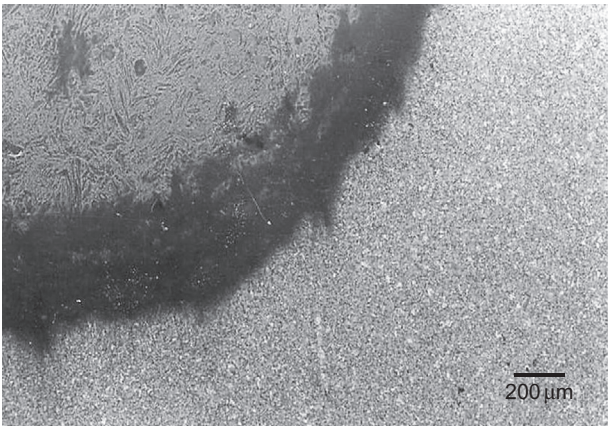


Fig. 5.4.8 Micrograph of a section of the unused drill rod sample showing sharp nicks on fillet contour at the collar-shank junction. Etchant: 2% nital.

(Fig. 5.4.9). The microstructure of the sample consisted of ferrite and lamellar pearlite (Fig. 5.4.10).

Hardness: Average Vickers hardness measured at 30kg load on the drill rod samples was as given here:

Sample	Hardness (HV)
Failed 1	440
2	390
Unused	395

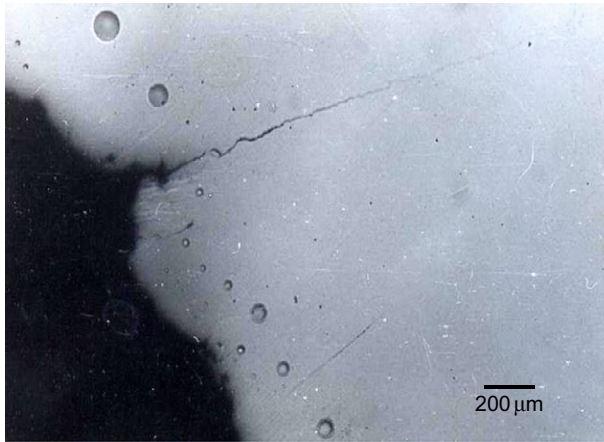


Fig. 5.4.9 Micrograph of a section of the unused drill rod sample showing transgranular crack propagating from one of the sharp nicks.

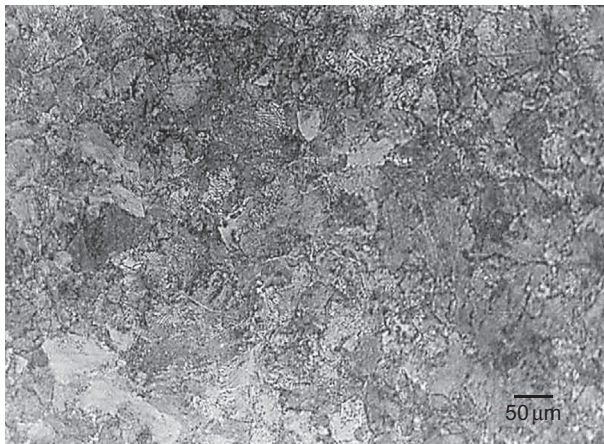


Fig. 5.4.10 Micrograph of a section of the unused drill rod sample showing ferrite and lamellar pearlite. Etchant: 2% nital.

5.4.3 Chemical Analysis (Wt.%)

Sample	C	Si	Mn	Cr	Mo	V	Cu	S	P	Fe
Failed	1.10	0.24	0.50	1.05	0.30	0.11	0.13	0.008	0.01	Rem.
Unused	1.10	0.29	0.48	0.96	0.28	0.10	0.11	0.006	0.01	Rem.

5.4.4 Discussion

Chemical analysis suggested that the drill rods have been made from 1C-1Cr-0.3Mo steel nearly equivalent to AISI grade L2 and conformed to the specification. Collar portion of both failed as well as the unused drill rod samples revealed a normalized microstructure. The hardness of the collar region of both failed and the unused samples was within the specified range. The collar-shank junction was seen to be rough with a sharp fillet radius and a number of sharp nicks were left on the surface. Crack near the collar-shank junction in the unused sample and oxides, pits, and cracks near the fracture origin in the failed samples clearly indicated that the cracks were preexisting in the drill rods which have probably resulted during the high temperature collar forging operation of the rods. Under the stress-raising effect of the preexisting cracks, the drill rods failed in a brittle cleavage mode during service. The normalized microstructure having poor impact toughness also had aided the low energy cracking of the rods. For a given strength level, a tempered martensite microstructure is known to impart the best combination of strength, ductility, and toughness for applications like that of drill rods. Hence hardening and tempering would have been a better heat treatment for the drill rods rather than normalizing.

5.4.5 Conclusion

The drill rods failed due to the presence of preexisting cracks near the collar-shank junction and an improper heat treatment of the collar portion resulting in a microstructure having poor impact toughness.

5.4.6 Recommendation

- (i) Exercise proper control over collar forging operation in order to avoid formation of cracks.
- (ii) Render higher fillet radius and proper surface finish at the collar-shank junction.
- (iii) Impart hardening and tempering treatment to the collar region instead of normalizing in order to obtain the specified hardness level.



5.5 CASE STUDY: FAILURE OF LIFTING LUGS OF A MISSILE

5.5.1 Background

Lifting lugs of a missile was found to contain a crack when it was taken up for use. The component was made using SAE 4340 grade steel in hardened and

tempered condition. The finished component was cadmium plated, hydrogen de-embrittled, and passivated before storage.

5.5.2 Experimental Results

Visual Examination: A cracked lug in the as-received condition is shown in Fig. 5.5.1. The lug showed deep grooves, steps, and severe roughness on the machined surface. The crack was forced open. The fracture surface showed dark color and rust across the stem and at least on half the section of the base (Fig. 5.5.2). Bright yellow color was also seen on one-third of the thickness on all four sides of the fracture surface at the base.

Fractographic Examination: The fracture surface was cleaned with oxide remover followed by acetone ultrasonically and examined under scanning electron microscope. The dark region of the fracture surface revealed predominantly an intergranular fracture with secondary cracks (Fig. 5.5.3). The remaining fracture surface showed fine dimpled rupture (Fig. 5.5.4).

Metallographic Examination: Longitudinal section of the lug showed elongated sulfide, globular oxide, and alumina inclusions when the specimen was examined under optical microscope. On etching with 2% nital the specimen

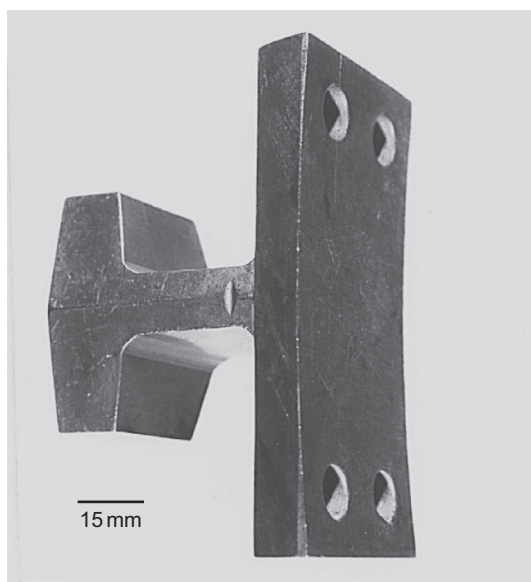


Fig. 5.5.1 A cracked lug in the as-received condition.

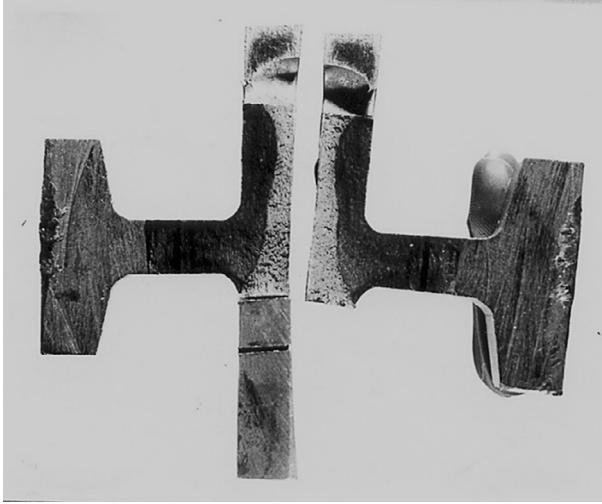


Fig. 5.5.2 The fracture surface showing dark color and rust across the stem and at least on half the section of the base.

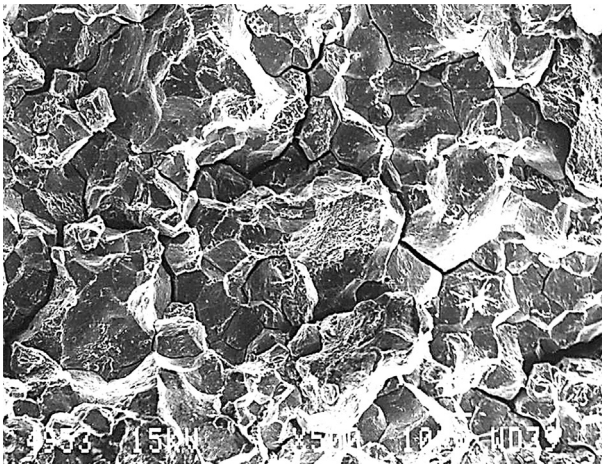


Fig. 5.5.3 SEM fractograph of the dark region of the fracture surface showing an intergranular fracture with secondary cracks.

revealed a tempered martensite microstructure. The outer surface of the specimen confirmed cadmium plating of about 50 μm .

Hardness: Average Vickers hardness of the lug was 372 HV at 30 kg load.

Chemical Composition: Chemical composition of the steel conformed to SAE 4340 grade.

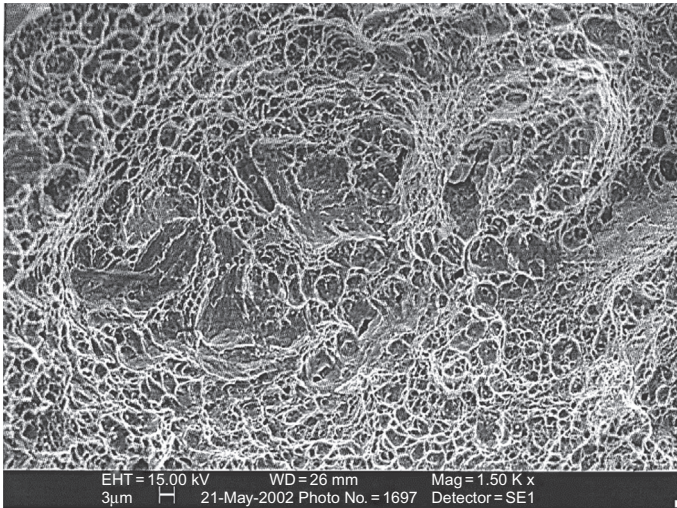


Fig. 5.5.4 SEM fractograph of the remaining part of the fracture surface showing fine equiaxed dimpled rupture.

5.5.3 Discussion

Chemical analysis, microstructure, and hardness indicated that the lug was made of the specified steel, i.e., SAE 4340 grade and used in a hardened and tempered condition. Absence of decarburized layer on the microstructure ruled out the possibility of crack existing prior to hardening treatment. Presence of bright yellow color on the fracture surface suggested that the crack existed during cadmium plating and so it was initiated at some stage of manufacture prior to cadmium plating. The dark color and rust observed on the fracture surface could have been caused by the contact of cracked inner surface with the plating and passivating bath solutions. The predominantly intergranular mode of the preexisting crack could be due to

- (a) Hydrogen embrittlement
- (b) Quench cracking

It is confirmed that the crack existed before cadmium plating and since hydrogen embrittlement often results after cadmium plating, this reason for cracking can be ruled out. Considering the significant variation in section thickness and poor surface finish after machining, the susceptibility of the lug to quench cracking is highly probable as these act as stress raisers.

5.5.4 Conclusion

From the present investigation it is reasonable to conclude that the lifting lug of the missile suffered quench cracking during hardening treatment. The quench cracking tendency developed in the component due to variation in section size and poor surface finish acting as stress raisers.



5.6 CASE STUDY: FAILURE OF BREECH BLOCK OF LIGHT MACHINE GUN

5.6.1 Background

A number of breechblocks of 5.56mm caliber light machine gun have failed prematurely during operation. The pressure developed inside the cartridge case during firing was 330–350 MPa. The breechblock keeps the cartridge in position inside the chamber during firing, retrieves the cartridge after firing, and also retains the firing pin. As it slides into the chamber and retracts continuously, it is designed for a high wear resistance in addition to other properties like strength, toughness, etc.

The breech blocks were made by machining from steel bar conforming to grade En 39B of BS 970—1955 and case carburized. Later they were hardened and tempered to a case hardness of 575–650 HV and a core hardness of 375–400 HV. The case-hardened surface was then finished by grinding to drawing dimensions.

5.6.2 Experimental Results

Visual Examination: One of the breech blocks showing a typical failure in the as-received condition is shown in Fig. 5.6.1. The breech block showed wide variation in section thickness. The failure was found to have occurred at the ejector chamber where the section thickness was the lowest. Cracks were also observed around the firing pin retainer hole adjacent to the extractor recess (Fig. 5.6.2).

Fractographic Examination: The fracture surface was cleaned and examined under SEM. It showed chevron pattern pointing to an origin located at the surface closer to the hole in the chamber. At fracture origin intergranular fracture features with secondary cracks were seen (Fig. 5.6.3). The rest of the fracture surface showed a mixed mode of quasi-cleavage and dimpled rupture (Fig. 5.6.4).

Metallographic Examination: Microscopic examination of a section of the breech block revealed a case depth varying from 0.25 mm to 0.60 mm on the

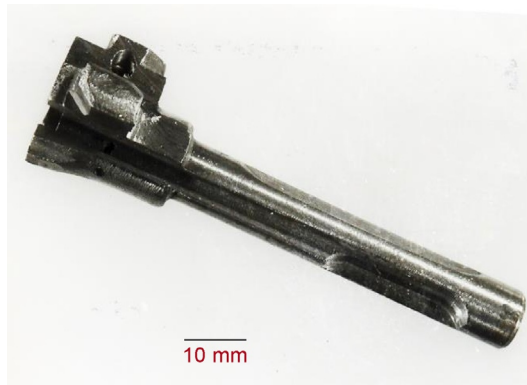


Fig. 5.6.1 Failed breech block in the as-received condition.



Fig. 5.6.2 Close-up view of the failed component showing cracks around the firing pin retainer hole adjacent to the extractor recess.

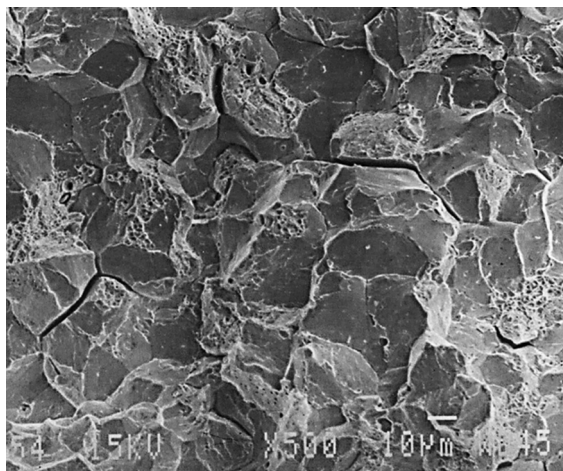


Fig. 5.6.3 SEM fractograph at the crack origin showing intergranular fracture with secondary cracks.

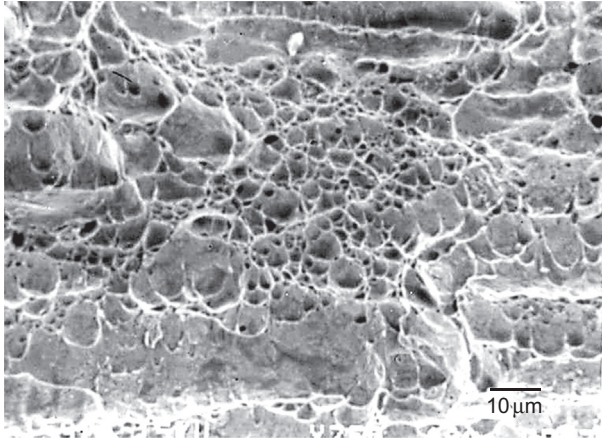


Fig. 5.6.4 SEM fractograph of the remaining fracture surface showing a mixed mode of quasi-cleavage and dimpled rupture.

surface leaving about an average of 0.40 mm thick core in the failed region (Fig. 5.6.5). Near the hole at one location, the entire section thickness showed a case (Fig. 5.6.6). The case showed acicular martensite and retained austenite (Figs. 5.6.5 and 5.6.6). The core of the sample showed a tempered martensite structure (Fig. 5.6.7).

Hardness: Micro hardness of the white etching area in the case measured 229 HV confirming that to be retained austenite. The average hardness of the case was 535 HV and that of the core was 490 HV.

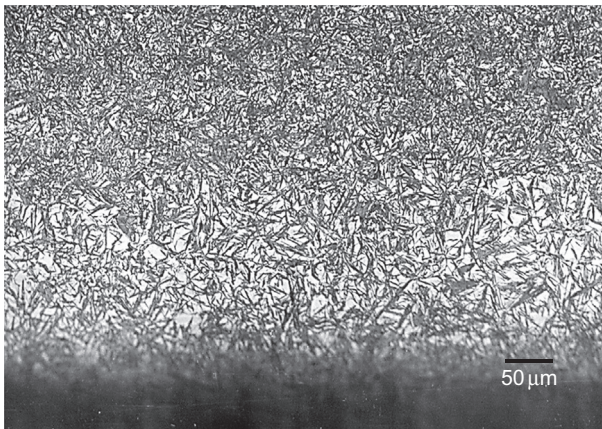


Fig. 5.6.5 Micrograph showing a deep case carburized layer near the failed region. Etchant: 2% nital.

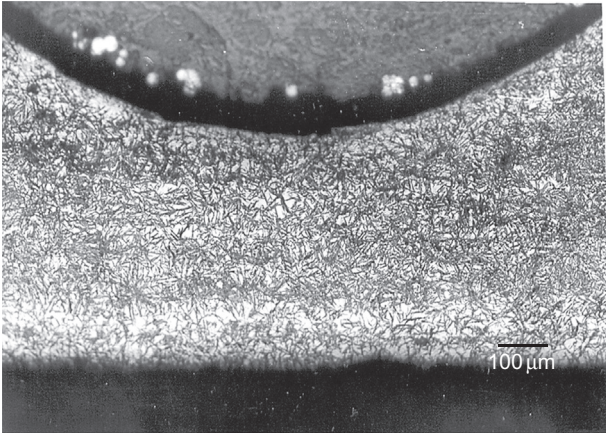


Fig. 5.6.6 Micrograph showing case carburized layer across the thickness near a hole. Etchant: 2% nital.

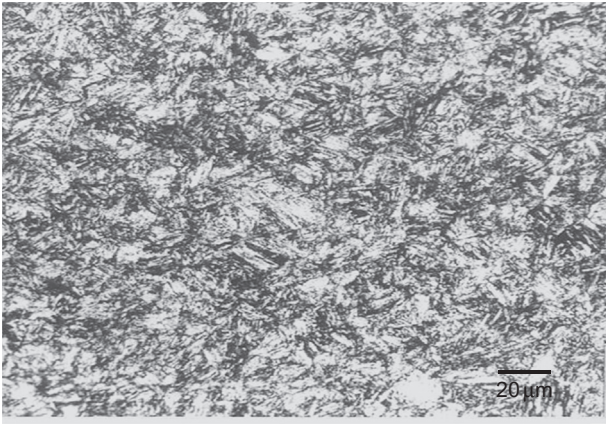


Fig. 5.6.7 Micrograph showing a tempered martensite structure at the core of the sample in the thick section. Etchant: 2% nital.

5.6.3 Chemical Analysis (Wt. %)

C	Si	Mn	Ni	Cr	Mo	S	P	Fe
0.106	0.23	0.27	4.30	0.86	0.22	0.018	0.012	Rem.

5.6.4 Discussion

The chemical composition of the breech block nearly conformed to the specified grade. In all the cases the fracture origins were located at the surface of ejector chamber that incidentally was case carburized, hardened, and finish ground. The case depth was much higher than that was required and so it left inadequate thickness of the core in the thinner sections. This was caused due to improper control over the case carburizing treatment. The presence of large quantity of retained austenite, which was confirmed by low hardness, was due to high carbon in the case. The retained austenite persisted in the case, as the tempering temperature and time were not high enough for its transformation. The higher core hardness than that was specified also indicated employment of lower tempering parameter after case hardening.

The fracture origin was located at the surface, which incidentally was case hardened and finish ground. Intergranular cracks had developed on the surface either during grinding of the case or during operation of the weapon. Under the influence of the stress-raising effect of the intergranular crack at the surface, the remaining crack had propagated in a mixed quasi-cleavage and dimple mode due to overload at a high strain rate.

5.6.5 Conclusion

The breechblock had failed due to improper control over case carburizing and heat treatment imparted to the component.

5.6.6 Recommendation

- (i) Exercise proper control over case carburizing process to obtain correct case depth especially in thin sections.
- (ii) Refrigeration treatment to be given soon after case hardening for transforming retained austenite.
- (iii) Use correct tempering parameter to achieve the required hardness and adequate toughness in the core.
- (iv) The possibility of using a through hardening steel and using in a hardened and tempered condition may be considered.



5.7 CASE STUDY: FAILURE OF TITANIUM SPONGE CUTTING RING

5.7.1 Background

A shredder system used for crushing titanium sponge lumps of about 50 mm diameter to a finished product size of 2–25 mm broke prematurely. The shredder system was designed by a private firm which consists of eight toothed rings of 30 mm thickness mounted horizontally on a shaft, rotating within a set of slots on a fixed breaker bar (called the comb). Each ring has two teeth diametrically opposite to each other. The rings are assembled on the shaft in such a way that the teeth have a staggered configuration; thereby the cutting action is restricted to one tooth at a time. The shredder operates at about 10 rpm. During trial crushing with an assembly of rings and comb, some of the rings broke into two halves prematurely. The rings and comb were made of Cr-Mo steel.

5.7.2 Experimental Results

Visual Examination: Broken part of one of the pieces of the ring in the as-received condition is shown in Fig. 5.7.1. The cutter ring was found to have broken from two portions that are almost diametrically opposite (A and B in Fig. 5.7.1). Fracture at location 'A' had initiated from the corner of the keyway and propagated radially outward at an angle of $\sim 0^\circ$ up to the section where the thickness decreases (Fig. 5.7.2). The fracture surface showed a coarse and crystalline fracture under stereomicroscope. Location 'A' showed



Fig. 5.7.1 The broken cutter ring in the as-received condition.

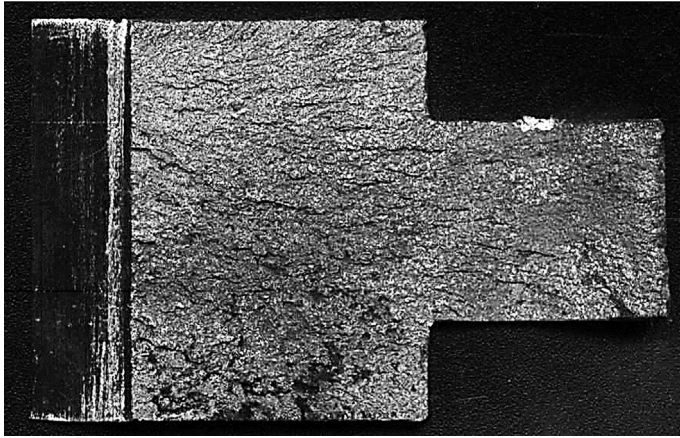


Fig. 5.7.2 Fracture surface at location 'A' showing a brittle fracture.

a fracture without any deformation except for a small shear lip at the final failure zone. Fracture at location 'B', on the other hand, revealed a shear at 45° originating at a corner of the cutting tool and propagating radially inward as seen by the chevron pattern on the fracture surface (Fig. 5.7.3). The fracture surface showed a fibrous fracture when examined under stereomicroscope.

Fractographic Examination: The fracture surfaces were cleaned and examined under scanning electron microscope. Location 'A' showed a quasi-cleavage fracture (Fig. 5.7.4), while location 'B' showed equiaxed dimples (Fig. 5.7.5).



Fig. 5.7.3 Fracture surface at location 'B' showing chevron marks.

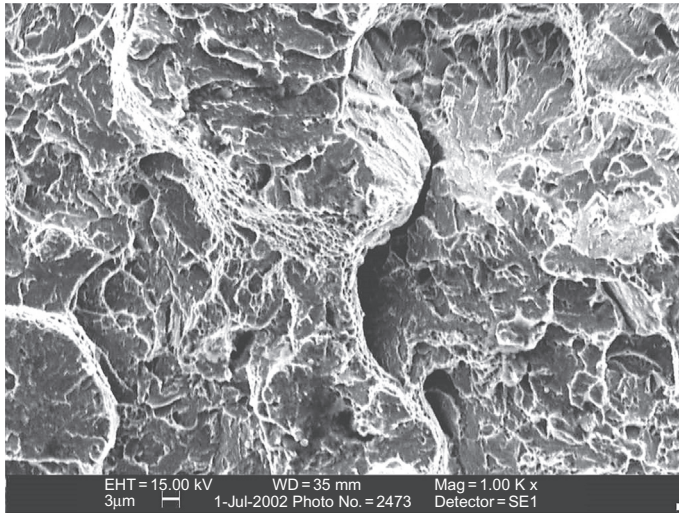


Fig. 5.7.4 SEM fractograph of the fracture surface at location 'A' showing a quasi-cleavage fracture.

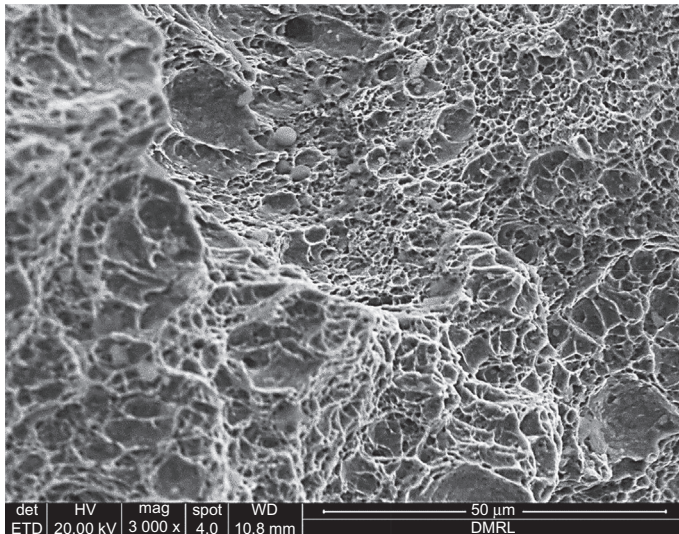


Fig. 5.7.5 SEM fractograph of the fracture surface at location 'B' showing a dimpled rupture.

Metallographic Examination: A section of the ring was prepared and examined under optical microscope. On etching with 2% nital, the sample revealed a cored structure of tempered martensite (Fig. 5.7.6).

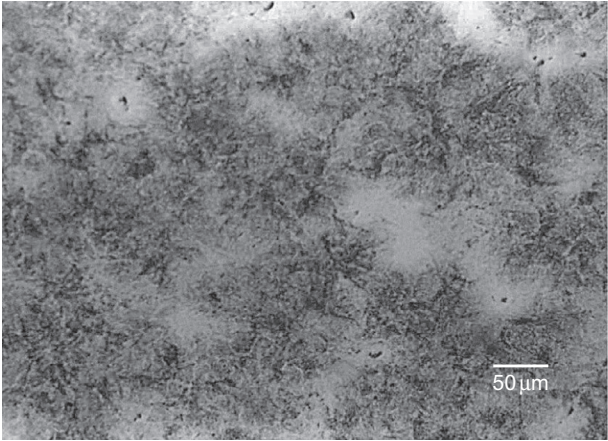


Fig. 5.7.6 Micrograph of a section of the cutter ring showing a cored structure of tempered martensite. Etchant: 2% nital.

Hardness: Average Vickers hardness of the ring measured at 30kg load was 537 HV.

5.7.3 Chemical Analysis (Wt. %)

C	Si	Mn	Cr	Mo	S	P	Fe
0.45	0.89	0.65	3.10	0.34	0.03	0.025	Base

5.7.4 Discussion

Chemical composition, microstructure, and hardness suggested that the failed cutter ring was a Cr-Mo steel casting used in hardened and tempered condition. Tool steels of this class are generally very hard in the as-hardened condition and exhibit a secondary hardening/retarded softening when tempered at about 500°C due to the formation of $(Cr, Mo)_x C_\gamma$ carbides. Hardness of the component was found to be quite high (51.5 HRC) which indicated that the casting was either tempered at low temperature or at secondary hardening range. Lower tempering temperature within the susceptible range can result in tempered martensite embrittlement (TME). Higher hardness and/or TME could cause reduction in both ductility and toughness of the casting. Same is the problem when tempered in the secondary hardening temperature range. Since the steel is not listed among the standard grades as per international specifications, the tempering behavior of the steel

needs a thorough investigation in order to select the most optimum tempering parameter taking into account the end use.

Visual examination showed that the failure had initiated from the corner of one of the keyways and propagated radially without much deformation. The design of the cutter ring showed that the cutting tooth of the cutter was along the key way. In the present design of the component the corner of the keyway can act as a stress raiser when a load is applied on the tooth during cutting/crushing operation. This can result in cracking at high strain rate because the toughness and the ductility of the casting were quite low to accommodate the stress as evidenced by quasi-cleavage fracture of the cutter ring at this portion. The mode of fracture also revealed that portion 'A' of the ring must have fractured first causing a condition of overload on portion 'B' and resulting in its failure as evidenced by fracture at an angle of about 45° in this area. The chevron marks and fibrous fracture features in this area also confirmed a ductile overload fracture.

5.7.5 Conclusion

The cutter ring had failed due to:

- (a) Stress concentration at the corner of the keyway causing a stress-raising effect, and
- (b) Very high hardness of the component resulting in lower ductility and toughness.

5.7.6 Remedial Measures

The following remedial measures are suggested for better performance of the cutter ring:

1. Change the design of the cutter ring so that the keyway and the cutting tooth are not placed in one line.
2. Exercise better control over casting method so as to produce a sound casting without any porosities and blow holes.
3. Heat treat the component to either a lower hardness level so as to render sufficient toughness and ductility to it that could withstand the applied load or impart a high case hardness with a softer core to absorb the impact load.



5.8 CASE STUDY: FAILURE OF COMBUSTION CHAMBER OUTER CASING OF AN AEROENGINE

5.8.1 Background

The combustion chamber outer casing of an aeroengine failed prematurely in service by tearing open a large area. The defect was suspected when the

hot gas was found leaking from the engine. The casing was stated to have been made from a nickel base alloy sheet metal. The casing was fitted on the hot zone of the engine surrounding the flame tube. The component is stated to be experiencing a maximum temperature of about 1200°C while in use and the cooling media was air. The component had completed 506:29 hours of service while the overhaul life was 1650 h.

5.8.2 Experimental Results

Visual Examination: The combustion chamber outer casing (C.C.O. casing) was observed to have failed along the weld between the elliptical area containing the three hollow cylindrical portions and the C.C.O casing sheet (Fig. 5.8.1). The fracture surface of a length of about 50 mm (arrow in Fig. 5.8.1) in the sheet was found to be smooth and dark with no noticeable deformation. The rest of the fracture surface was bright and fibrous with a 45° shear failure.

Fractographic Examination: Sample was cut from smooth and dark area of the fracture surface. The fracture surface was cleaned ultrasonically and examined under scanning electron microscope (SEM). Fracture surface near the inner surface of the casing contained a number of cavities (Fig. 5.8.2). Cleave-like steps were seen originating from large cavities near the inner surface of the casing (Fig. 5.8.3). Characteristics fatigue striations were seen superimposed on the cleave steps (Fig. 5.8.4). The striations were masked by spherical pits (Fig. 5.8.4). The final failure of less than one-fourth of the fracture surface showed dimpled rupture (Fig. 5.8.5).

Metallographic Examination: Samples were cut from near the failed portion of the casing and prepared for metallographic examination.

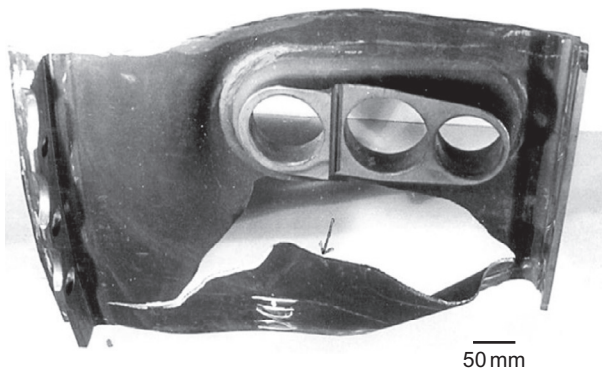


Fig. 5.8.1 The failed part of CCO casing in the as-received condition.

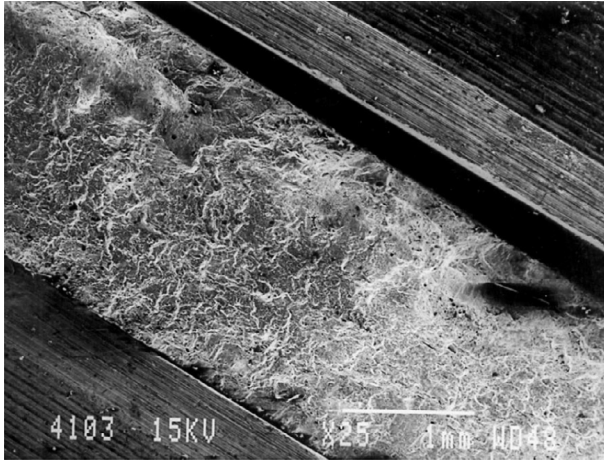


Fig. 5.8.2 SEM fractograph of the fracture surface showing a number of cavities near the inner surface of the casing.

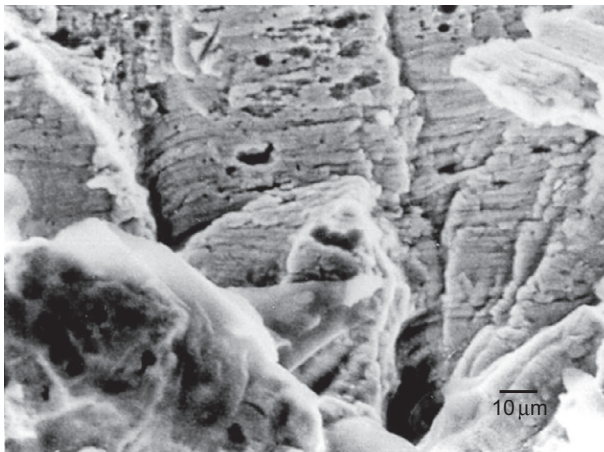


Fig. 5.8.3 SEM fractograph of the fracture surface showing cleave-like step originating from large cavities near the inner surface of the casing.

In the unetched condition, the sample showed a number of oxide and nitride inclusions. On etching with acidified ferric chloride the sample revealed failure along the weld (Fig. 5.8.6). The sheet sample showed coarse equiaxed grains of austenite with annealing twins containing carbides and nitrides (Fig. 5.8.7)

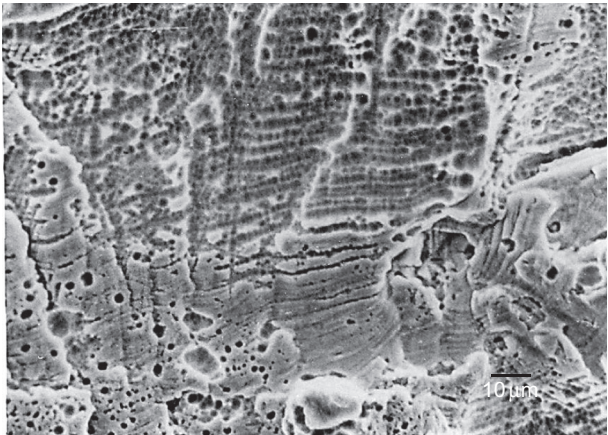


Fig. 5.8.4 SEM fractograph of the fracture surface showing characteristic fatigue striations masked by globular pits.

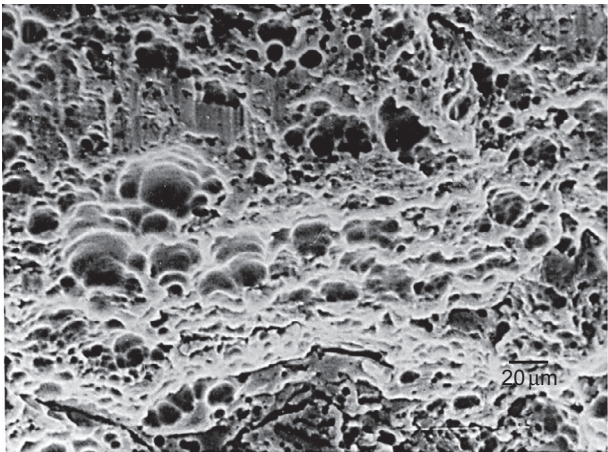


Fig. 5.8.5 SEM fractograph of the fracture surface showing dimpled rupture in the final failure zone.

Hardness: Hardness of the C.C.O. casing was measured on the sheet across the weld at 10kg load and the results are given here.

Location	Hardness (HV)
Casing sheet	315
Heat affected zone	275
Weld	254
Elliptical region	224

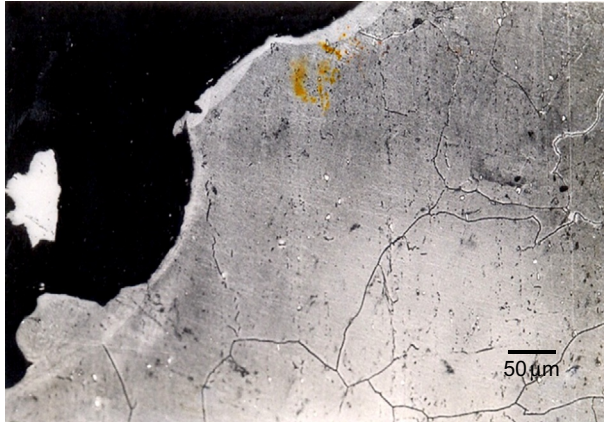


Fig. 5.8.6 Micrograph of a section of the sample showing primary cracking along the weld. Etchant: Acidified ferric chloride.

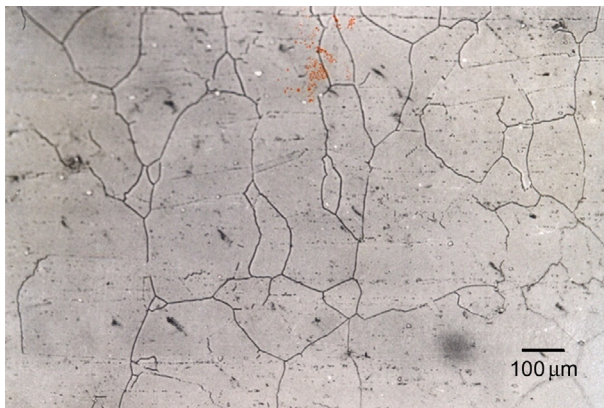


Fig. 5.8.7 Micrograph of a section of the sample showing coarse equiaxed grains of austenite with annealing twins containing carbides and nitrides. Etchant: Acidified ferric chloride.

5.8.3 Chemical Analysis (Wt.%)

Cr	Ni	C	Mn	Si	Fe
24.30	17.90	0.15	1.20	0.45	Base

5.8.4 Discussion

Chemical composition suggested that the combustion chamber outer casing was made using an austenitic stainless steel of grade 20kh23N18 of GOST 5632-72. The steel was used in an annealed condition. The higher hardness of the casing sheet was due to the cold work done on it during its fabrication and use. The coarse grains of the casing were due to its prolonged operation at high temperature.

The weld between the elliptical portion containing three cylindrical holes and the casing sheet revealed large cavities near the inner surface indicating defects in the weld. Fractographic examination clearly indicated the fracture origin to be at these weld defects. The crack had propagated under fatigue loading conditions as evidenced by fatigue striations in SEM fractography. Evidences of pits on the fracture surface indicated that the fracture surface was exposed to very high temperatures during service. The steel is recommended for high temperature applications up to 1000°C for the prolonged period. However, the onset of intensive scaling is at 1050°C in this steel [1]. The high scale pitting seen on the fracture surface could be due to its exposure to temperatures above 1050°C during operation. The 50 mm long crack through the thickness of the sheet at the center near the weld joint must have existed for some time before the final tearing of the 200 mm long crack that resulted due to overload. This is evidenced by the dark color of the original fracture surface and scale on it. The stress concentration at the ends of the initial crack must have caused the crack propagation on both sides resulting in an overload failure of the casing.

5.8.5 Conclusion

Combustion chamber outer casing has failed due to fatigue initiated at weld defects near the inner surface. The final failure was due to overload.

5.8.6 Reference

- [1] GOST 5632-72, grade 20kh23N18, State Standards of the USSR, British Standards Institution, Linford Wood, Milton Keynes, 1982.



Casting Defects in Iron and Steel

Imperfections and discontinuities can occur in iron and steel castings. If they go undetected, they can initiate failures during service. Some of the commonly found defects in casting are: cold shut, gas porosity, hot tear, shrinkage cavity, inclusions, seam, stress cracks, etc. Although a sound foundry practice can reduce occurrence of casting defects, nondestructive testing of castings should be mandatory for detection of defects in order to avoid entry of defective casting into the assembly.

(a) *Segregation*: Segregation is understood as the variation in chemical composition of a solidified casting from its mean composition. This defect is generally associated with alloys. As alloys consist of more than one component element, the solidification of an alloy usually occurs over a range of temperatures. As the liquid solidifies the solid phase that is formed first with drop in temperature bears a different composition than the mean composition of the alloy. This results in rejection of solute from the solidifying phase into the liquid that would solidify at a lower temperature. This kind of solute rejection continues with decrease in temperature until the last solid is formed. Thus the solidification of an alloy results in large variation in chemical composition in the casting causing segregation.

Macro etching of a section is a simple technique that reveals existence of segregation in a cast product. Differential etching of the segregated area leaves a pattern on the etched surface. The pattern forms a qualitative index of the amount of segregation. Electron probe microanalysis (EPMA) is an accurate method for quantitative analysis of segregation. The method is described in Section 3.9.

Segregation in a casting can affect corrosion resistance, mechanical properties, and variation in response to heat treatment from point to point, the net result would be a catastrophic failure of the casting during its service.

The chemical inhomogeneity caused by segregation can be removed from a casting or an ingot by homogenizing treatment for a long time at a suitable temperature.

(b) *Blow Holes and Porosity*: Entrapment of gases in a solidifying metal leaves blowholes and porosity in a casting as defects. These defects may occur as discrete round holes or in the form of clusters.

Radiography of castings is a standard technique that reveals the presence of large blowholes and porosity. When these defects are larger in number and finer in size, they are detected by optical microscopic examination. There are standard techniques mentioned in international specifications by which these types of defects in castings can be identified and quantified.

Presence of blowholes and porosity primarily reduces the effective cross-sectional area of the sound metal and thereby the load-carrying capability of the casting. They can aid in fatigue crack propagation and reduce the corrosion resistance of the casting as well.

(c) *Shrinkage*: During solidification of a casting the surface cools and solidifies first as it is in contact with the atmosphere. The inside of the casting is still hot due to the molten metal, which is not in direct contact with the atmosphere. Thus the solidification of a casting starts at the surface and proceeds inward. On similar argument it can be easily understood that thinner section gets solidified first as compared to the thicker section of a casting. Contraction of a metal is a simultaneous process with cooling and solidification of molten metal. Thus the metal that solidifies last, which is usually near the core of the casting experiences the maximum contraction resulting in the formation of a shrinkage cavity. The size of the shrinkage cavity is related to the amount of metal feeding from a molten metal pool, the temperature of casting, and the volume changes occurring during solidification.

Since shrinkage cavities are basically internal voids in an otherwise sound looking casting, it can be detected easily by X-ray radiography. Massive shrinkage cavities as well as cavities occupying larger volumes are readily detected by radiography. Small cavities at grain boundaries and interdendritic cavities are detected by metallography.

(d) *Pipe*: Most metals and alloys contract on solidification to about 4% by volume. As described earlier, the last metal to solidify during ingot solidification results in formation of a shrinkage cavity at the center and top of the ingot known as 'pipe'. The pipe may extend deeper into the ingot, which may be detected as secondary pipe and porosity.

Radiography detects the primary pipe that is generally cropped before the ingot is taken up for further processing. However, if secondary pipe or porosity goes undetected in a radiography it can lead to centerline defects in wrought products. Removal of primary pipe by cropping is essential as the exposed surface on top of the ingot gets oxidized and therefore fails to weld during subsequent forging and rolling resulting in defects in the processed alloy. Cropping of the pipe portion of an ingot although cause loss of metal and affect economy, it can avoid further loss by way of a failure.

(e) *Hot Tears*: Crack-like defects, which are generally discontinuous appear either on the surface of a casting or internally are called 'hot tears'. This is again associated with differential contraction of varying sections in a casting. Sections cooling faster can pull away from sections cooling slowly due to temperature gradients resulting in hot tears.

Internal tears can be detected by radiography and ultrasonic inspection. Being crack-like defect, the hot tears can propagate a crack under service conditions in a component.

(f) *Inclusions*: Inclusions are oxides, sulfides, and nitrides of metals. Since their properties are different from the bulk metal, they are considered as nonmetallic inclusions. Inclusions can get into the metal from particles of refractory from the furnace lining, ladle and molds during melting and casting. Deoxidation carried out using aluminum and silicon during steel making also can leave some residual oxide inclusions in the metal. Nitrogen from air or from raw materials can form nitrides of metals that readily combine with it. Sulfides form from sulfur content in the raw materials. The kind of inclusions formed in the casting or ingot, their size, shape and distribution in the as-cast product and after processing of the ingot influence the mechanical properties and corrosion resistance of the component. Inclusions and their effects are described in Chapter 4.



6.1 CASE STUDY: FAILURE OF EXHAUST OUTER CONE OF AN AEROENGINE

6.1.1 Background

An aeroengine was received from one of the operating units with the defect of oil leak from the joint of combustion chamber outer casing and exhaust cone. On strip examination of the aeroengine a crack of about 50–75 mm was observed on the front flange of exhaust outer cone at 6-O'clock position in front of combustion drain valve. It is reported that the maximum working temperature of the part was 800°C. The number of hours completed by the part was also not reported although it was stated that the part's total technical life is assessed based on its condition. A section of the failed component containing the cracked portion was received for failure analysis.

6.1.2 Experimental Results

Visual Examination: A portion of the exhaust outer cone in the as-received condition is shown in Fig. 6.1.1. From the portion it can be visualized that the exhaust outer cone is made of a sheet metal that has been welded to the



Fig. 6.1.1 Portion of the exhaust outer cone in the as-received condition.

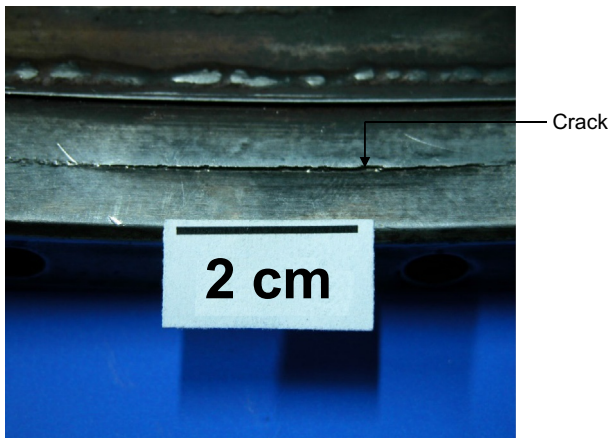


Fig. 6.1.2 Close-up of the front flange showing the crack.

front flange. A crack of about 60 mm length was observed on the front flange (Fig. 6.1.2). The crack was developed on the arm of the 'L' shaped flange which is welded to the cone. The section thickness of the flange on this arm was about 1.25 mm.

Fractographic Examination: The crack was cut open and cleaned ultrasonically in acetone and examined under SEM. About two-thirds of the section was coarse and crystalline with a number of cavities. The remaining one-

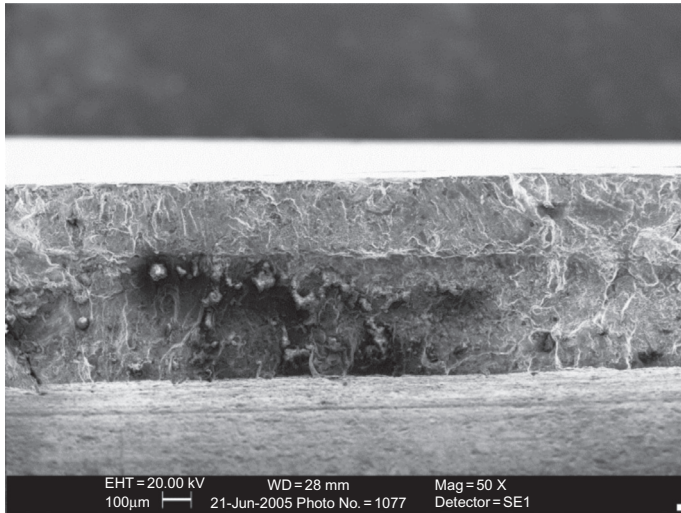


Fig. 6.1.3 SEM fractograph of the fracture surface showing about two-third of the fracture being coarse with large cavities and inclusions, and almost smooth fracture of the remaining one-third portion.

third of the fracture surface showed a mixed mode of smooth and coarse fracture (Fig. 6.1.3). Large cavities with some inclusions were noticed in the coarse region of the fracture surface. One such cavity containing some inclusions is shown in Fig. 6.1.4. The inclusions at two different spots were

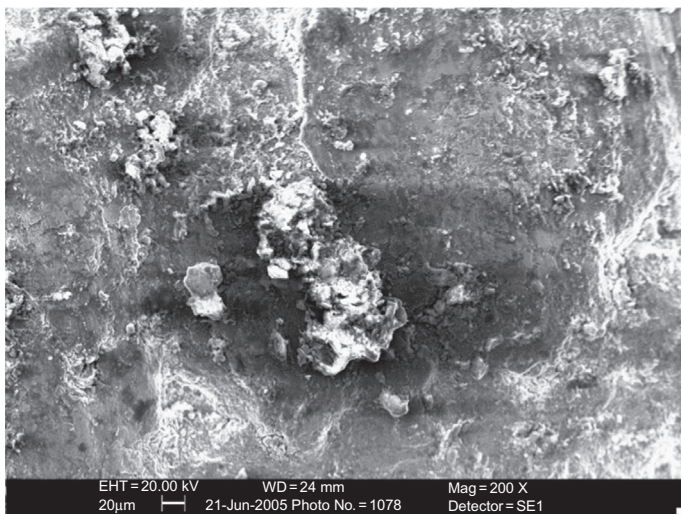


Fig. 6.1.4 SEM fractograph of the fracture surface showing one of the cavities containing some inclusions.

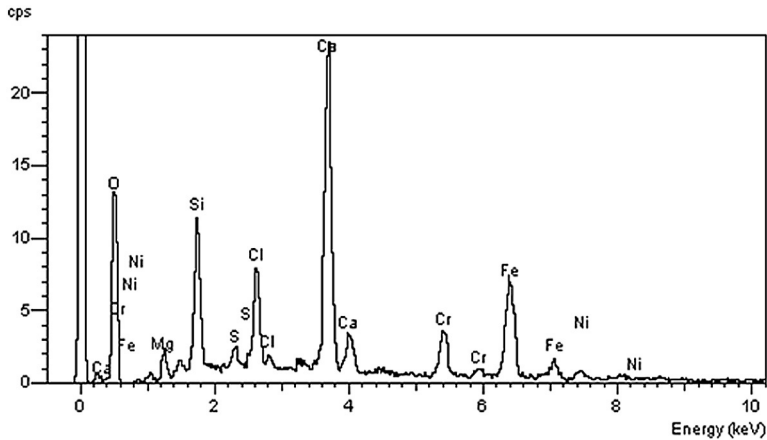


Fig. 6.1.5 EDAX spectrum of inclusions showing the presence of Si, Ca, Mg, Cl, O, and S in addition to Cr, Ni, and Fe.

analyzed by energy dispersive X-ray spectroscopy (EDAX) and found to contain Si, Mg, Ca, Cl, S, and O in addition to Cr, Ni, and Fe (Fig. 6.1.5). The coarse region of the fracture surface revealed a quasi-cleavage fracture (Fig. 6.1.6). The smooth area beyond the coarse fracture area showed characteristic fatigue striations (Fig. 6.1.7).

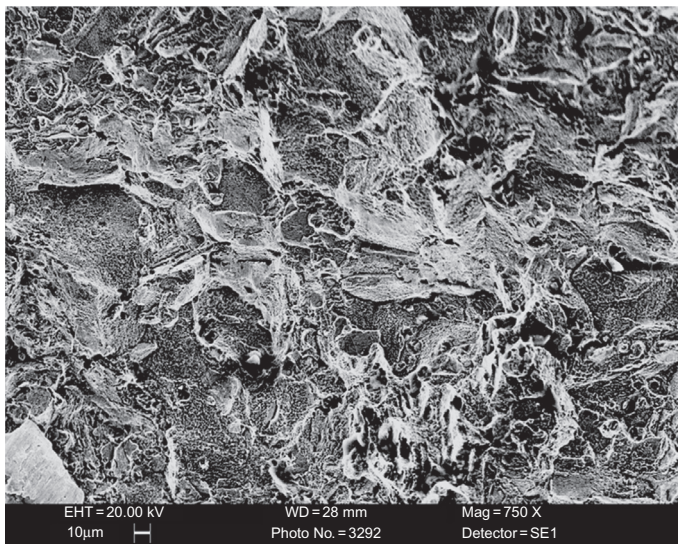


Fig. 6.1.6 SEM fractograph of the coarse region of the fracture surface showing quasi-cleavage fracture.

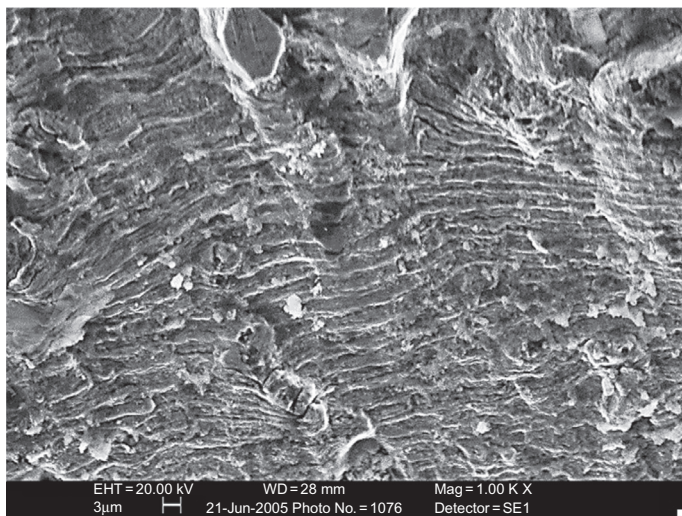


Fig. 6.1.7 SEM fractograph of the smooth region of the fracture surface showing fatigue striations.

Metallographic Examination: One section of the exhaust outer cone sample containing the thicker arm (on which the holes were drilled) and the thinner arm (which was welded to the cone portion) was mounted and prepared for metallography. In the unetched condition, the flange portion showed a number of globular oxide inclusions. On etching with aqua regia the flange portion of the sample revealed a cast structure (Fig. 6.1.8). The dendritic microstructure of the casting showed a number of interdendritic microcavities. The cone portion of the sample showed equiaxed grains of austenite with annealing twins (Fig. 6.1.9).

Hardness: Hardness was measured on the flange portion of the sample using a 1/16 inch ball indenter and Rockwell ‘B’ Scale. The average hardness was 85 HRB which is equivalent to 165HV.

6.1.3 Chemical Analysis of the Flange (Wt. %)

C	Si	Mn	Cr	Ni	Nb	S	Fe
0.072	1.05	0.79	18.00	8.56	0.50	0.008	Base

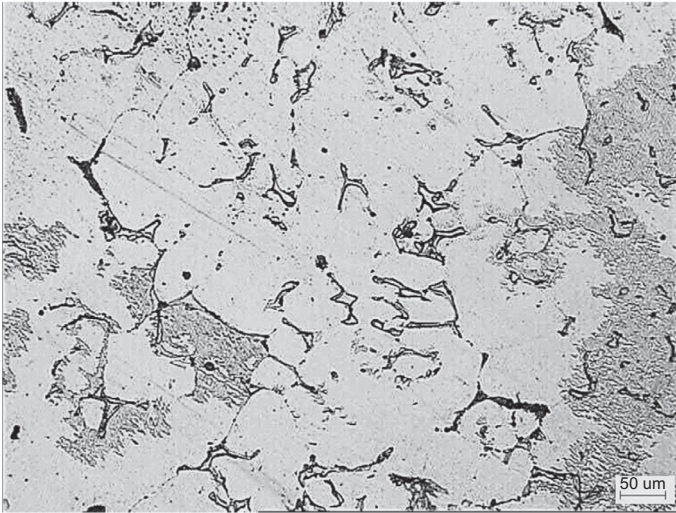


Fig. 6.1.8 Micrograph of the flange sample showing a dendritic microstructure and interdendritic microcavities. Etchant: Aqua regia.



Fig. 6.1.9 Micrograph near the weld region of the sample showing equiaxed grains of austenite with annealing twins in the cone portion. Etchant: Aqua regia.

6.1.4 Discussion

Chemical analysis and microstructure suggested that the flange portion of the exhaust cone was an austenitic stainless steel casting nearly conforming to UNS-J92710/ACI-CF8C used in the as-cast condition. The casting contained a number of defects viz., microcavities, porosities, and slag inclusions.

The defective portion of the flange occupied about two-thirds of the section thickness from where the crack originated as seen by SEM fractographic examination. Casting defects can reduce the effective cross-sectional area especially in lower cross sections of components. Lower cross sections will not be able to withstand the designed load as load-bearing capacity decreases which will in turn have an adverse effect on the fatigue strength of the component.

The thicker arm (containing the drilled holes) of the flange was about 3mm thick while the thinner arm which was welded to the cone portion was about 1.5mm thick. The lower thickness of the thinner arm of the flange was probably designed so as to match with the section of the cone portion with which it is joined by welding. If maintaining such thin section of this arm of the flange was a design requirement, then manufacturing it by casting can be considered as an inadequate design. Moreover, the control over the casting method was very poor because of which the casting contained a large number of interdendritic microcavities and slag pockets. The cavities and slag pockets over about two-thirds of the section thickness reduced the effective section size as well as acted as stress raisers which initiated the fatigue crack. The crack had initiated from the center and propagated on both sides circumferentially under fatigue-loading condition as evidenced in fractography.

6.1.5 Conclusion

The front flange of the exhaust outer cone developed a crack due to casting defects viz., microcavities, porosities, and slag inclusions which propagated under fatigue-loading conditions over a distance of about 60mm in length and entire section thickness resulting in a through and through crack. The oil leakage was due to this cracking of the front flange in the exhaust outer cone of the aeroengine.

6.1.6 Recommendations

1. If casting is unavoidable it should be carried out with utmost control so that no casting defects are left in the component.
2. The front flange should be subjected to radiography for checking casting defects before assembly. The NDT check should also be incorporated in the quality line during maintenance and overhaul for the presence of crack-like defects.

3. Closed die forging is considered a better manufacturing method for components of lower thicknesses like that of the front flange of the exhaust outer cone as compared to casting; in order to produce a defect-free product.



6.2 CASE STUDY: FAILURE OF SEALING RINGS OF CENTRIFUGAL BREATHER OF AN AEROENGINE

6.2.1 Background

After five hours of operation, engine oil leakage through centrifugal breather was observed from the aeroengine of a helicopter. On examination it was found that sealing rings fitted on the centrifugal breather were broken, the purpose of which was to prevent oil leakage from centrifugal breather gear. The oil used was OX-38/Castrol 98. Maximum temperature of the oil reaches 70°C with a maximum pressure of 4 kg/cm² during operation of the aeroengine. Two failed sealing rings from the centrifugal breather gear were taken for investigation. The samples are identified as A and B.

6.2.2 Experimental Results

Visual Examination: The sealing rings in the as-received condition are shown in Fig. 6.2.1. The fracture surface when examined under stereomicroscope was seen to be coarse and crystalline.

Fractographic Examination: On examining under SEM the fracture surface of sample (A) revealed transgranular cleavage fracture in the broken sealing ring while the sample (B) showed quasi-cleavage fracture (Figs. 6.2.2 and 6.2.3).

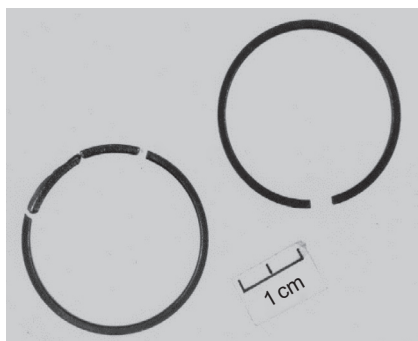


Fig. 6.2.1 Failed sealing rings in the as-received condition.

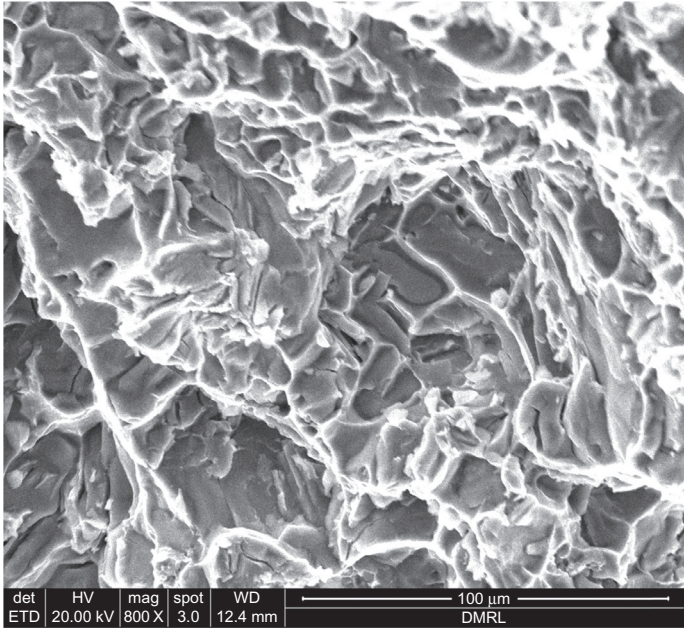


Fig. 6.2.2 SEM fractograph of the fracture surface showing transgranular cleavage fracture in sample A.

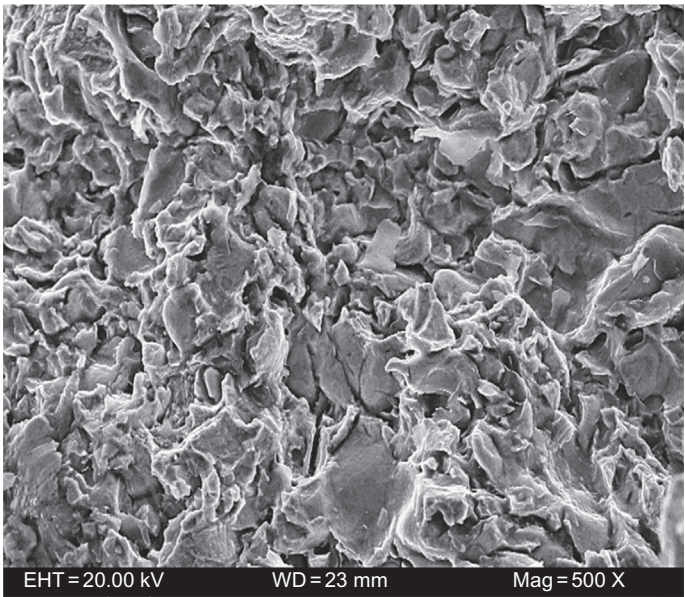


Fig. 6.2.3 SEM fractograph of the fracture surface showing quasi-cleavage fracture in sample B.

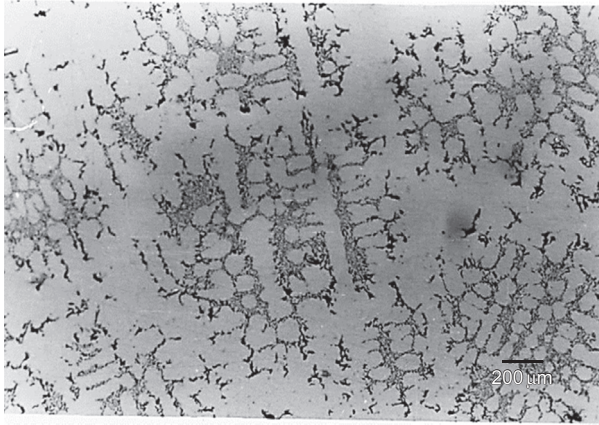


Fig. 6.2.4 A typical micrograph showing E-type graphite distribution in a section of sample A. Unetched.

Metallographic Examination: Sections of the two failed samples were prepared for metallographic examination. In the unetched condition, sample (A) showed interdendritic segregation of fine graphite flakes having preferred orientation i.e., E-type distribution as per ASTM A247 (Fig. 6.2.4) [1]. In the unetched condition, sample (B) showed random distribution of graphite flakes and rosette grouping, i.e., a mixed A-B type distribution (Fig. 6.2.5) [1]. On etching with 2% nital, the failed sample (A) revealed interdendritic segregation of finer graphite flakes with preferred orientation in a matrix of lamellar pearlite with free cementite (Fig. 6.2.6). On etching

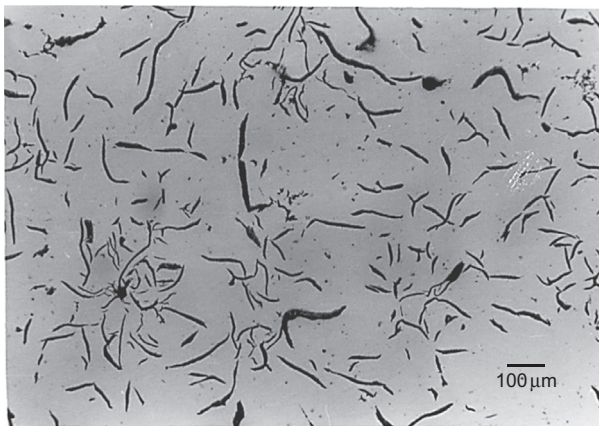


Fig. 6.2.5 A typical micrograph showing mixed A-B type graphite distribution in a section of sample B. Unetched.

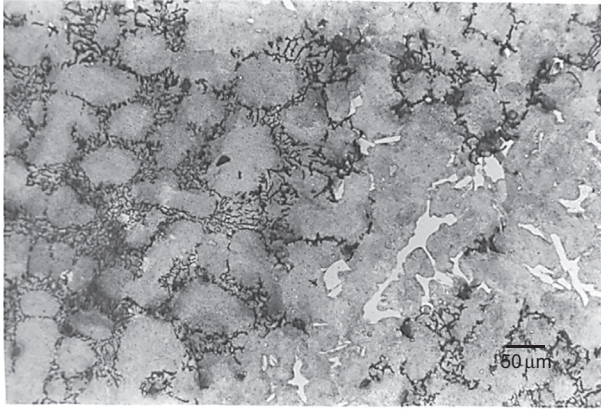


Fig. 6.2.6 Micrograph of a section of sample A revealing interdendritic segregation of finer graphite flakes with preferred orientation in a matrix of lamellar pearlite with free cementite. Etchant: 2% nital.

with 2% nital, the sample (B) showed lamellar pearlite and islands of free carbides along with A-B type graphite distribution (Fig. 6.2.7). Metallographic examination of the samples used for SEM fractography revealed that the sample failing in cleavage mode was corresponding to E-type distribution of graphite, whereas the sample failing in quasi-cleavage mode was corresponding to A-B type graphite distribution.

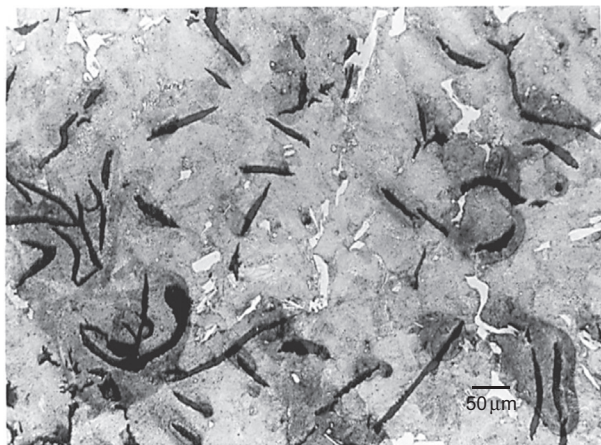


Fig. 6.2.7 Micrograph of a section of sample B revealing lamellar pearlite and islands of free carbides along with A-B type graphite distribution. Etchant: 2% nital.

Hardness: Average Vickers hardness of the samples at 5 kg load was as given here:

Sample A: 290 HV, Sample B: 265 HV.

6.2.3 Chemical Analysis (Wt. %)

Sample	C	Si	Mn	Cr	Mo	S	P	Fe
A	3.20	2.09	0.70	1.08	0.64	0.02	0.20	Base
B	3.10	2.34	0.78	1.05	0.65	0.02	0.24	Base

6.2.4 Discussion

Chemical compositions of the failed sealing rings suggested that they were cast using low alloy cast iron nearly conforming to grade 1 of IS: 7925-1976 (Bureau of Indian Standards) and were used in the as-cast condition. However, the failed samples showed two types of graphite distribution in the microstructure. Fractography revealed that the microstructure of the cast iron influenced its fracture mode. Sealing ring with E-type graphite distribution had failed in transgranular cleavage mode as compared to A-B type graphite distribution that exhibited quasi-cleavage fracture. The interdendritic graphite flakes as seen in sample A are undesirable when compared with random distribution of graphite flakes and rosette grouping as seen in sample B from the point of view of toughness of the cast iron. Very high volume fraction of free carbide in the microstructure also affects the ductility, a common feature that was observed in both the failed samples. Energy absorbed during fracture by cleavage mode should have been lower than quasi-cleavage fracture mode. It can be inferred that the sealing ring sample with E-type graphite distribution was more brittle as compared to sealing rings with A-B type graphite distribution. However, for most of the engineering applications graphite distributions of type A (as per ASTM A247) are preferred over the others.

Thus from the investigation, it is evident that the microstructure of the cast iron influences brittle fracture. Both E type and A-B type of graphite distribution caused failure of sealing rings. It is therefore necessary to control both the graphite morphology and distribution and also the free carbide by sound foundry practice. Inoculation, composition control, and cooling rate are the important parameters, which have to be judiciously controlled to achieve the desirable type A graphite distribution in pearlite matrix.

6.2.5 Conclusion

The sealing rings had failed due to undesirable microstructure consisting of E type and A-B type graphite distribution that influenced brittle fracture of the rings.

6.2.6 Reference

- [1] ASTM Standard A247-98, Section 01, Vol. 1.02.



6.3 CASE STUDY: FAILURE OF GRINDING ROLL OF A BOWL MILL

6.3.1 Background

Grinding rolls used in XRP.783 type bowl mill of a thermal power station failed prematurely. The grinding rolls weighing about one ton each were made using Ni-hard cast iron. The grinding rolls were expected to have a service life of 4000 h while they had to be withdrawn from service after 2100 h due to extensive abrasion of the working surface. As per the specification, the grinding rolls were to be centrifugally cast from Ni-hard material and suitably heat treated to a minimum hardness of 580 HB.

A small portion of failed grinding roll weighing about 1.5 kg was received for failure analysis.

6.3.2 Experimental Results

Visual Examination: The sample piece of the failed grinding roll in the as-received condition is shown in Fig. 6.3.1. One of the surfaces of the sample corresponded with the undamaged surface of the roll while the working surface showed a wavy pattern due to abrasion. The abraded surface was spoilt by sputtered metal due to gas cutting and so the surface could not be examined minutely.

Metallographic Examination: Sections of the roll sample were cut away from the gas cut end by EDM wire cutting method that contained the original and the abraded surfaces and prepared for metallographic examination. In the unetched condition, the sample showed a large number of shrinkage cavities throughout the polished surface (Fig. 6.3.2). On etching with 2% nital, the sample revealed massive carbide in a matrix of fine pearlite, which is typical of a white cast iron (Figs. 6.3.3 and 6.3.4).

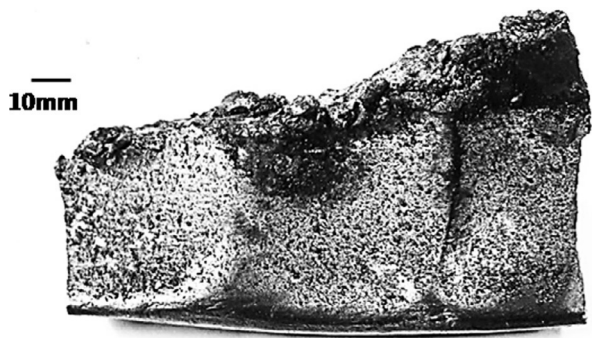


Fig. 6.3.1 Sample piece from the failed grinding roll in the as-received condition.

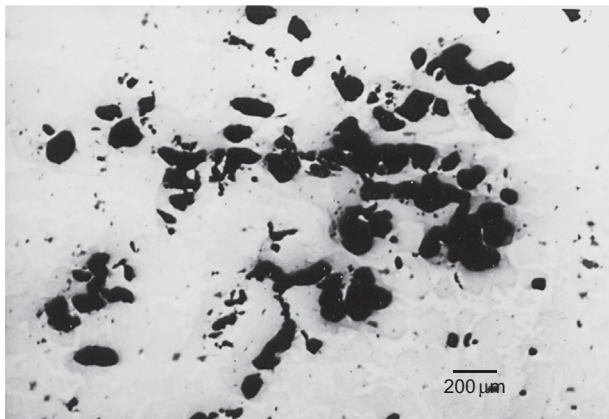


Fig. 6.3.2 Micrograph of a section of the failed roll showing a large number of shrinkage cavities. Unetched.

Hardness: Brinell hardness was measured using a 10mm diameter hardened steel ball at 3000 kg load and the average hardness was 415 HB.

6.3.3 Chemical Analysis (Wt.%)

C	S	P	Mn	Si	Ni	Cr	Fe
2.77	0.16	0.30	1.00	1.20	4.50	3.00	Balance

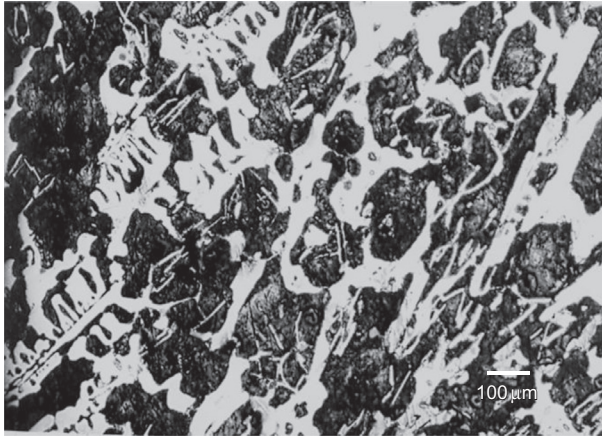


Fig. 6.3.3 Micrograph of a section of the failed roll showing massive carbide in a matrix of fine pearlite. Etchant: 2% nital.

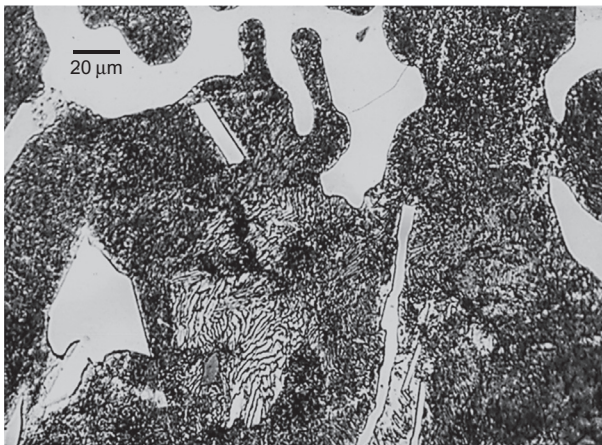


Fig. 6.3.4 Same as [Fig. 6.3.3](#) at higher magnification. Note the lamellar nature of pearlite phase. Etchant: 2% nital.

6.3.4 Discussion

Chemical analysis suggested that the grinding roll was made of a cast iron that was nearly conforming to Ni-hard type II. The desired microstructure of Ni-hard grinding roll was massive carbides in a matrix of martensite containing a small amount of retained austenite. Such a microstructure would possess a very high hardness (~ 650 HB) that is extremely desirable for imparting necessary abrasion resistance on the working surface of the

grinding rolls. The prematurely failed grinding roll revealed a microstructure containing massive carbide in a matrix of pearlite. The hardness of the roll was much lower than the specified minimum.

The presence of fine pearlite in the microstructure indicated that the grinding roll was not cooled sufficiently rapidly below the eutectoid temperature resulting in the formation of softer constituents. Had the cooling rate been sufficiently rapid, the eutectoid austenite would have missed the pearlite transformation bay and thus would have converted into martensite below the M_s temperature. It appeared that the grinding roll was cast without using proper chill. The presence of large amount of shrinkage cavities indicated that proper feeding arrangements were not used in order to eliminate such defects during solidification of the casting. Presence of such casting defects also contribute to lower life of the grinding roll.

6.3.5 Conclusion

The premature failure of the grinding roll was due to improper microstructure and lower hardness than that was specified for the purpose.

6.3.6 Recommendation

1. The pearlitic structure of the grinding roll cannot be altered to martensitic structure by heat treatment without causing any damage to the roll.
2. Avoid using the grinding rolls from the same manufacturing lot, which would give lower service life than that is required.



Overload

Under service conditions a component may be subjected to tensile, bend, shear, compressive, torsion, or impact stresses. In certain situations, these stresses may cross the limit of design stress causing overload and subsequent failure of the component. Overload failure can also occur if the material strength is much lower that does not allow it to withstand the applied load. Sometimes it is difficult for the designer to accurately predict the actual service stresses. Metallurgical defects in the material such as inclusions, cavities, porosities, blow holes, etc. may reduce the effective cross-sectional area and can result in overload failure. Degradation of the material due to embrittlement or corrosion of the part can lead to reduction in load-bearing capability of the component and resulting in an overload failure in service. Improper design, material selection, or heat treatment can all play a part in causing overload failure.

An overload failure is manifested both in ductile or brittle mode depending on various factors that include the characteristics of the material as well as the environment. Overload failures occurring at ambient temperature and high temperature leave their respective tell-tale evidences on the fracture surface.

Factor of safety is the ratio of yield strength of the material to design load of the part. A design engineer always plans for a high factor of safety in order to accommodate for an operator error or an unforeseen shock load that the part might be subjected to during its operation. For example, a factor of safety of 10 is used in fasteners. When a factor of safety is used in the design, it should always be based upon the yield strength of material. Often designers calculate the factor of safety based on ultimate tensile strength, which is a design error and can be a cause of overload failure.

Macroscopically an overload failure shows gross deformation of a ductile material near the failure. Thinning, necking, buckling, bending, elongation, formation of shear lip, etc. that shows gross dimensional change of the failed part are tell-tale evidences of ductile overload failure. An intrinsically brittle material or a material that was embrittled due to metallurgical or environmental reasons, presence of a notch or notch-like defect, etc., when failed

due to overload or an impact overload, do not exhibit any deformation near the failure. Under scanning electron microscope, a ductile overload failure reveals equiaxed or elongated dimples, whereas a brittle overload failure reveals a transgranular cleavage or a quasi-cleavage fracture feature. In case of high temperature overload failures, the fracture surface generally shows intergranular fracture features.



7.1 CASE STUDY: FAILURE OF MOUNTING BOLT OF EMERGENCY CHANGEOVER VALVE

7.1.1 Background

There was a case of precautionary landing of a helicopter due to failure of main hydraulic system. During post incident inspection, the mounting bolt of emergency change over valve was found broken. Emergency change over valve is a mechanical component that allows main hydraulic system to function when the hydraulic pressure is between 45 ± 3 and $65 \pm 2 \text{ kg/cm}^2$. There is a duplicate hydraulic system that maintains the pressure and holds against the spring tension. The moment the main pressure drops below $30 \pm 5 \text{ kg/cm}^2$ the duplicate hydraulic system takes over. Emergency change over valve is attached to cradle of helicopter by two mounting bolts. Close-up view showing the mounting bolt fitted on the emergency change over valve is presented in Fig. 7.1.1.

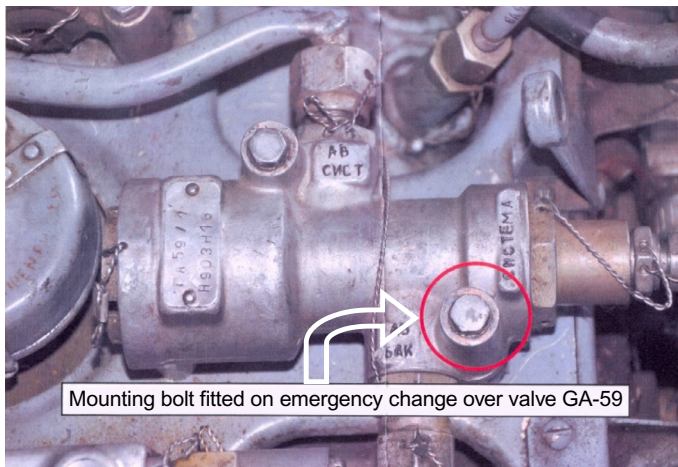


Fig. 7.1.1 Close-up view showing the mounting bolt fitted on the emergency change over valve.

7.1.2 Experimental Results

Visual Examination: The failed mounting bolt in the as-received condition is shown in Fig. 7.1.2. The bolt was found to have broken from the threaded region.

Fractographic Examination: Both the mating fracture surfaces of the bolt were cleaned ultrasonically in acetone and examined under scanning electron microscope (SEM). Both the mating fracture surfaces of the bolt showed fibrous fracture features (Figs. 7.1.3 and 7.1.4). One of the fracture surfaces showed severe rub marks (Fig. 7.1.4). At higher magnification the fracture surfaces revealed slightly elongated dimples (Fig. 7.1.5).

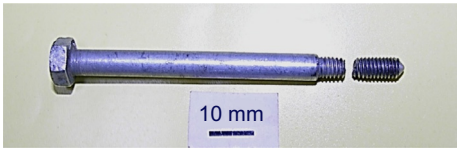


Fig. 7.1.2 Failed bolt in the as-received condition.

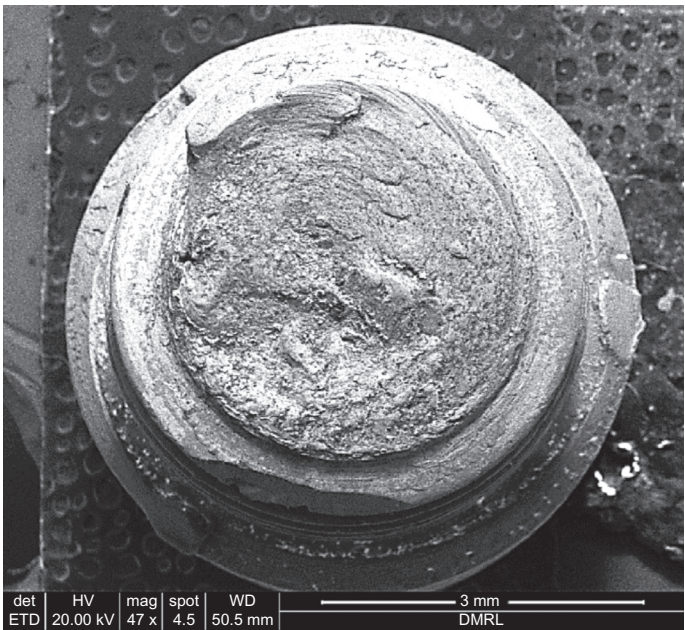


Fig. 7.1.3 Fracture surface of the bolt showing a fibrous fracture feature.

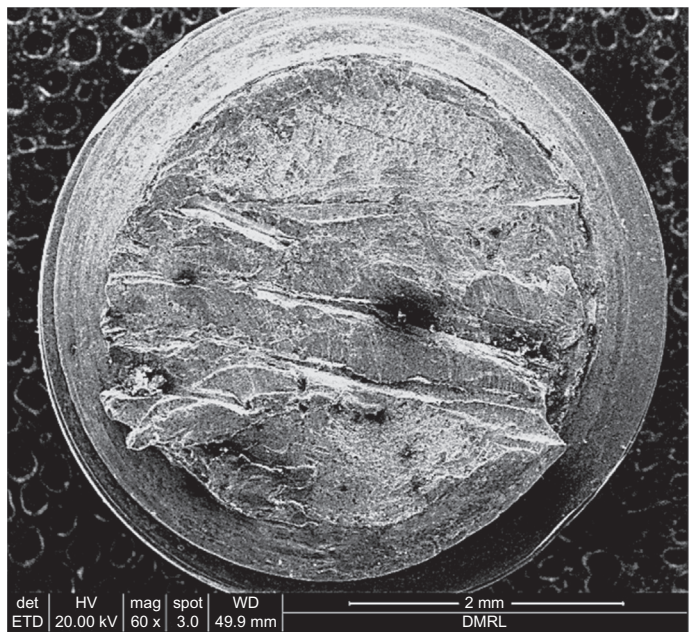


Fig. 7.1.4 Another mating fracture surface of the bolt showing a fibrous fracture feature and severe rub marks.

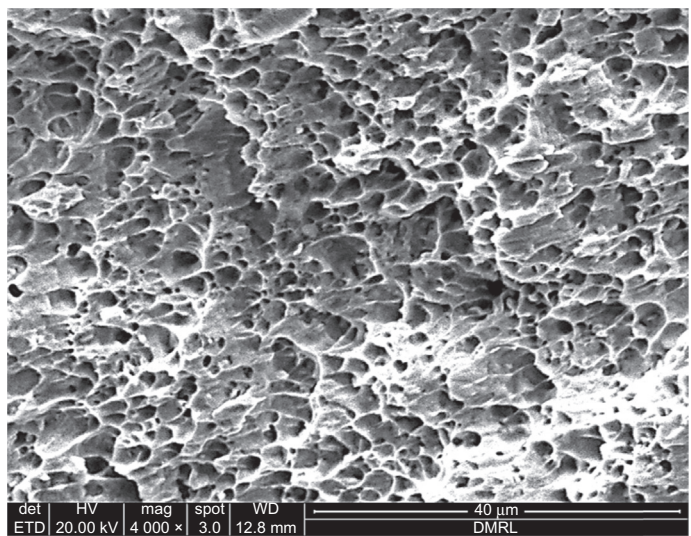


Fig. 7.1.5 SEM fractograph of the fracture surface showing slightly elongated dimples.

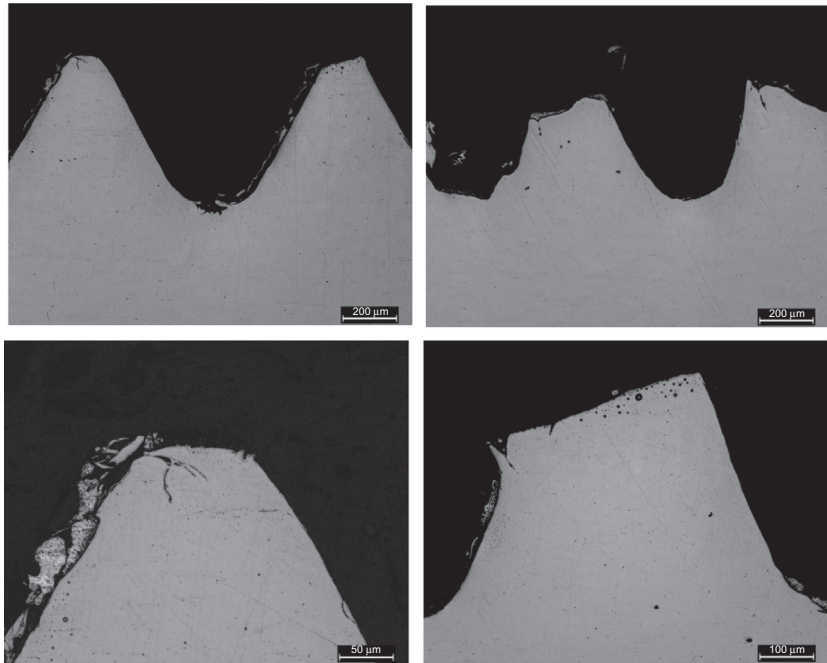


Fig. 7.1.6 Micrographs of longitudinal section of the failed bolt showing cracks both at the crest and trough of the threads and deformation of the threaded region. Unetched.

Metallographic Examination: Longitudinal section of the bolt sample was prepared for metallographic examination. In the unetched condition the section showed cracks both at the crest as well as at the trough of the threads (Fig. 7.1.6). Material deformation was also noticed in the threaded region (Fig. 7.1.6). The sample showed a number of thick elongated sulfides and fine globular oxides (Fig. 7.1.7). On etching with 2% nital, the sample revealed spheroidized carbides in a matrix of ferrite (Fig. 7.1.8).

Hardness: Average Vickers hardness of the bolt measured at 10 kg load was 200 HV.

7.1.3 Chemical Analysis (Wt.%)

C	Si	Mn	Cr	S	Fe
0.47	0.56	0.70	0.88	0.020	Balance

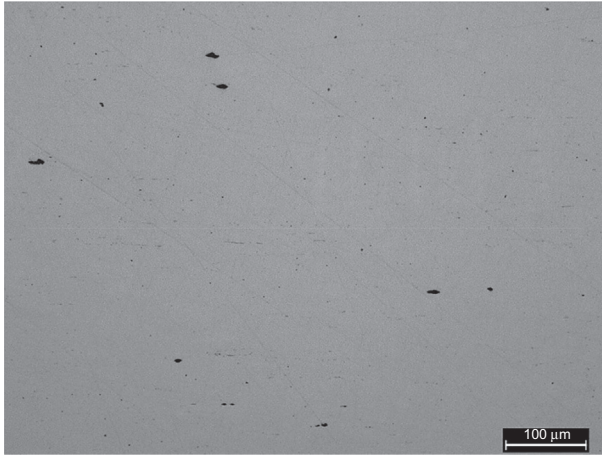


Fig. 7.1.7 Micrograph of the longitudinal section of the failed bolt showing elongated sulfide and globular oxide inclusions. Unetched.

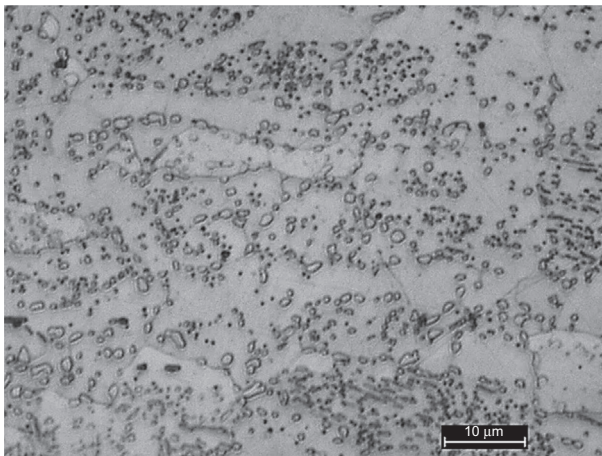


Fig. 7.1.8 Micrograph of the longitudinal section of the failed bolt showing spheroidized carbides in a matrix of ferrite. Etchant: 2% nital.

7.1.4 Discussion

Chemical composition of the failed bolt suggested that it was made of a low alloy steel nearly conforming to grade 50X of GOST 4543-71 (Russian specification). Microstructure and hardness indicated that the bolt was used in a spheroidized annealed condition.

The steel used, being a medium carbon low alloy steel, is generally used in a hardened and tempered condition that provides required strength and adequate toughness for the intended application. However, from the micro-structure and hardness value it is obvious that the steel was not used in the required heat-treated condition in order to develop optimum mechanical properties. Spheroidized annealed condition is relatively soft that would not develop much strength in the material.

Fractographic examination confirmed the failure to be due to overload. The bolt must have been subjected to overload either during fixing in the tapped hole or in service.

7.1.5 Conclusion

- (i) The mounting bolt had failed due to overload which could be the result of possessing lower strength.
- (ii) The bolt was used in spheroidized annealed condition instead of hardened and tempered condition which developed lower strength probably inadequate for the intended application.



7.2 CASE STUDY: FAILURE OF STEAM PIPE IN UNIT-VI OF A THERMAL POWER STATION

7.2.1 Background

There was a case of premature failure of steam pipe in unit-VI of a thermal power station. The pipe was made of a carbon steel of grade SA 106B (ASME specification). The failed pipe was a part of flow nozzle assembly. The nozzle was used to measure the steam flow at a rated pressure and temperature of 11 kg/cm^2 and 390°C respectively in the pipe line.

In the pipe line assembly steam enters at a temperature of 520°C and a pressure of 139 kg/cm^2 . By the time the steam passes from zone 1 to zone 2, the temperature drops to 475°C and the pressure drops to 14.4 kg/cm^2 . A water spray is admitted before the flow nozzle in the second zone. The flow nozzle is located in the third zone of the pipe line. The pressure and temperature drops to 11 kg/cm^2 and 390°C , respectively, in this zone.

7.2.2 Experimental Results

Visual Examination: The burst pipe in the as-received condition is shown in Fig. 7.2.1. The pipe was found to have flared and deformed near the fracture. One side of the inner surface of the pipe showed machined grooves which



Fig. 7.2.1 The burst pipe in the as-received condition.



Fig. 7.2.2 Close up of the inner side of the failed part of the pipe showing deep-machined grooves especially at the junction of the grooving and the smooth region.

was very deep at the junction of the grooved region and the smooth region below (Fig. 7.2.2). The pipe had burst with a 45° shear all along the fracture surface (Fig. 7.2.3). The fracture surface was examined using a magnifying glass. It did not show any origin of failure. Therefore, a number of samples of the fracture surface were taken for further examination by scanning electron microscope (SEM). The dimensions of the burst pipe were as given here.

Mean diameter (OD)—Away from failure: 275 mm, near failure: 300 mm

Thickness (W)—Away from failure: 10 mm, near failure: 8 mm

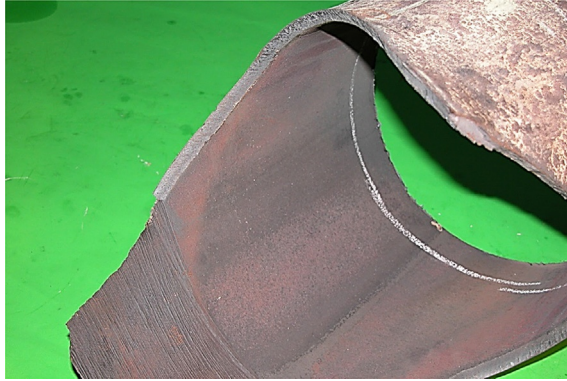


Fig. 7.2.3 A large portion of the fracture surface of the burst pipe showing a 45° shear.

Fractographic Examination: The samples from the fracture surface were cleaned ultrasonically in acetone and examined under SEM. The fracture surface revealed dimpled rupture in all the samples. A typical fractograph is shown in [Fig. 7.2.4](#).

Metallographic Examination: Longitudinal sections of samples were taken from different locations of the failed pipe and prepared for metallographic examination. In the unetched condition, the pipe samples showed a number of elongated sulfide and globular oxide inclusions. On etching with 2% nital,

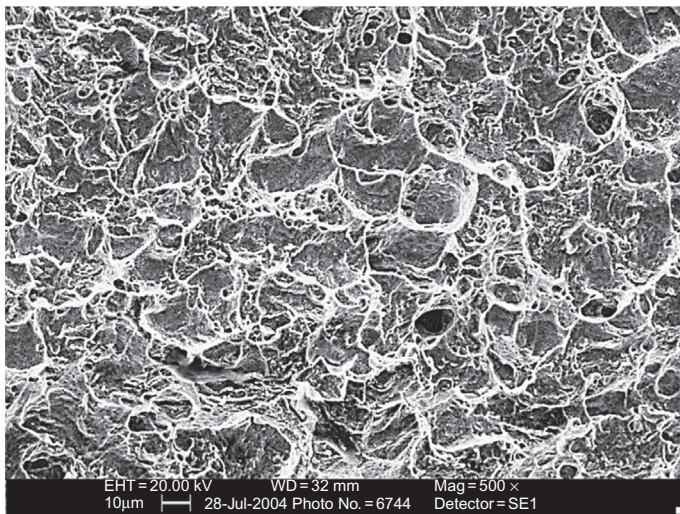


Fig. 7.2.4 SEM fractograph of a typical fracture surface showing dimpled rupture.

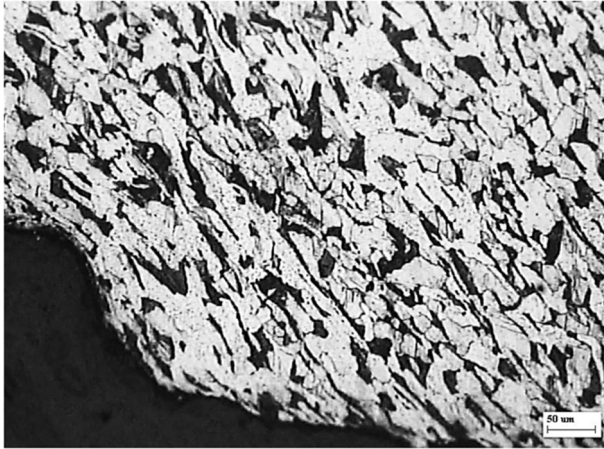


Fig. 7.2.5 Micrograph of a section of the sample taken from a region near the fracture showing elongated grains of ferrite and pearlite. Etchant: 2% nital.

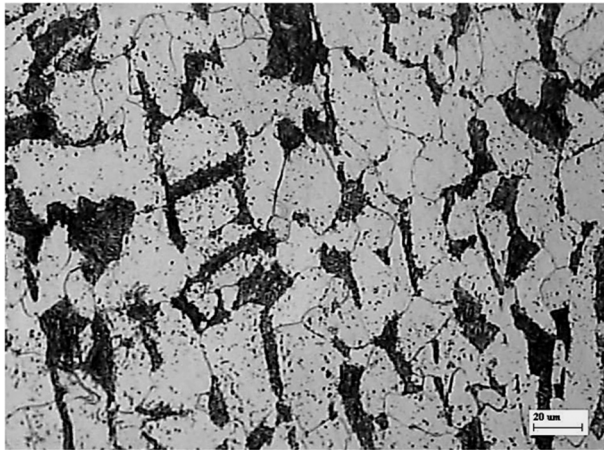


Fig. 7.2.6 Micrograph of the sample as described in [Fig. 7.2.5](#) showing pearlite (dark) and spheroidized carbides within the ferrite phase. Etchant: 2% nital.

the sample taken from an area near the fracture showed slip bands and elongated grains of ferrite and pearlite ([Fig. 7.2.5](#)). On observing the section at higher magnification it revealed spheroidized carbides within the ferrite phase and colonies of pearlite ([Fig. 7.2.6](#)). The carbides were found to be segregated more near the pearlite colony as though they were formed out of pearlite phase. The sample away from the fracture showed equiaxed grains of ferrite and colonies of pearlite ([Fig. 7.2.7](#)). The pearlite colony showed lamellar structure.

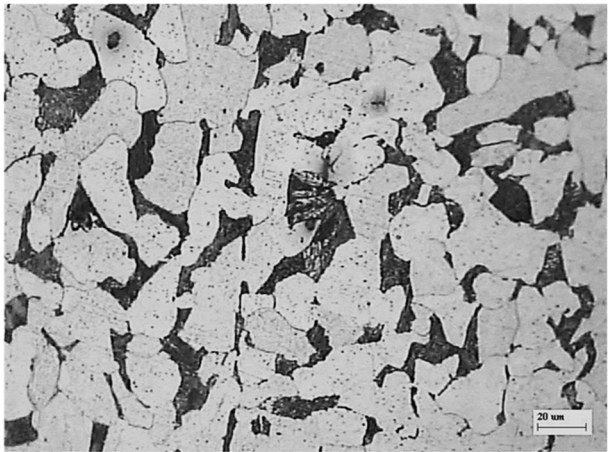


Fig. 7.2.7 Micrograph of a section of the sample taken from a region away from the fracture showing equiaxed grains of ferrite and colonies of pearlite. Etchant: 2 % nital.

7.2.3 Chemical Analysis (Wt.%)

C	Si	Mn	S	P	Fe
0.20	0.28	0.77	0.01	0.015	Base

7.2.4 Mechanical Properties

Tensile Properties (at Room Temperature)
Grade SA-106B ASME Specification [1]

Yield Strength	241 MPa (2457 kg/cm ²)
UTS	415 MPa (4232 kg/cm ²)
Elongation (50 mm gage length)	30%

Hardness: The average Vickers hardness values of the samples taken from different locations and measured at 20 kg load were as given here.

Location	Hardness (HV)
Near the fracture	231
Near machined surface	200
Flared region	187
Away from failure	158

Stress Calculation: A closed pipe subjected to certain internal pressure behaves as a pressure vessel wherein the major component of stress is the Hoop stress.

Hoop stress for a cylinder with certain internal pressure is given by the following relation:

$$S = PD_m/2W \quad [2]$$

where S is the Hoop stress (in kg/mm^2), P is the internal pressure (in kg/mm^2), D_m is the mean diameter of the pipe (in mm) and is equal to $(D - W)$, D is the outer diameter of the pipe (in mm), and W is the wall thickness (in mm).

An estimate of the original Hoop stress when the pipe was new is made from the dimensions of the pipe away from failure as given here:

$$P = 0.11 \text{ kg/mm}^2, D = 275 \text{ mm}, W = 10.0 \text{ mm}$$

$$S = \frac{0.11 \times (275 - 10)}{2 \times 10} = 1.46 \text{ kg/mm}^2 \quad (146 \text{ kg/cm}^2)$$

An estimate of the Hoop stress just before failure is made from the dimensions of the pipe near the failure as given here:

$$P = 0.11 \text{ kg/mm}^2, D = 300 \text{ mm}, W = 8.0 \text{ mm}$$

$$S = \frac{0.11 \times (300 - 8.0)}{2 \times 8.0} = 2.00 \text{ kg/mm}^2 \quad (200 \text{ kg/cm}^2)$$

7.2.5 Discussion

Chemical analysis suggested that the burst pipe was manufactured from a carbon steel grade SA-106B (ASME Specification) which was conforming to the specified material. Microstructure and hardness of the pipe away from failure indicated that the pipe was used in annealed condition. The pipe is expected to have adequate strength at the designed operating temperature and pressure (390°C and 11 kg/cm^2 , respectively) in annealed condition. Progressive increase in hardness of the pipe from intact region to the fractured region suggested cold working of the pipe before final failure. Presence of spheroidized carbides in the microstructure points to the fact that the pipe was heated to a temperature just below the AC_1 before the failure. However, partial transformation of pearlite into spheroidal carbides indicated that the pipe was heated locally for a relatively shorter duration of time. Development of higher pressure and temperature than the stipulated ones in the pipe led to flaring (bulging) of the pipe resulting in some amount of thinning from 10mm to 8 mm as noted from the thickness of the pipe in regions away and

near to the failure. Deep grooving due to machining of a part of the pipe could act as a stress raiser causing initiation of the crack.

Local flaring of the pipe had taken place after the partial transformation of the microstructure from a ferrite-pearlite structure to a ferrite-pearlite-spheroidized carbide structure, as evidenced by the elongated grains near the failure, which also confirmed the pipe temperature to be a little below the AC_1 in the failed region. From the ASME boiler and pressure vessel code it is gathered that the maximum allowable stress level for grade SA-106 B steel at 399°C is 9.16 kg/mm^2 (916 kg/cm^2), and at 538°C is 1.76 kg/mm^2 (176 kg/cm^2) [3]. As there was no evidence of creep it is inferred that the overheating to certain high temperature (but below AC_1) of the pipe was for a relatively short duration. During this period the hoop stress developed caused further flaring of the pipe until the stress crossed the maximum allowable stress level resulting in overload failure of the pipe.

7.2.6 Conclusion

High temperature in the range of $500\text{--}650^\circ\text{C}$ and high stress condition developed in the steam pipeline probably due to some fault in the maintenance has led to an overload failure of the pipe used in the flow nozzle assembly. Deep grooving on the inner surface of the pipe could have aided in crack initiation.

7.2.7 References

- [1] Metals Handbook, vol. 1, 9th ed., ASM, 1978, P648.
- [2] Metals Handbook, vol. 11, 9th ed., ASM, 1986, P611.
- [3] ASME Boiler and Pressure Vessel Code Table UCS-23, Section VIII, Division 1, 1986 Edition, p. 274.



7.3 CASE STUDY: FAILURE OF BODY OF SADDLE ASSEMBLY

7.3.1 Background

During the structural qualification testing of corset the body of the saddle assembly failed. It is stated that the failed component was made from AL-4 grade aluminum alloy casting imported and issued to a private firm for corset assembly.

Corset is meant for transportation and storage of the canister with or without the missile. It experiences various combinations of loads during

transportation by rail, road, and air. Two corsets used for railway transportation had to be tested for mechanical load in order to validate the corsets for the given loads. Compressive loads were applied in steps in all the three directions, held the set up under load for 10 min, and released gradually. It was observed that the corset withstood the applied loads, without any visual failure in case-1. However, in case-2, the body of the saddle of Corset-II was found broken.

7.3.2 Experimental Results

Visual Examination: The failed component in the as-received condition is shown in Fig. 7.3.1. The fracture surfaced did not show any probable crack origin. The fracture surfaces showed a mixed mode fracture i.e. a mixture of fibrous and granular fracture (Fig. 7.3.2).

Fractographic Examination: The fracture surfaces were cleaned ultrasonically in acetone and examined under scanning electron microscope (SEM). The fracture surfaces that did not reveal any explicit origin predominantly revealed a mixed mode of fracture comprising quasi-cleavage and dimpled rupture (Figs. 7.3.3 and 7.3.4).

Metallographic Examination: Sections in all the three directions of the components were prepared for metallographic examination. In the unetched condition, the sample showed a few interdendritic cavities and gas porosities (Fig. 7.3.5). On etching with Keller's reagent the samples revealed a cast structure consisting of an interdendritic network of eutectic silicon (gray) and some particles of Mg_2Si (black) with dendrites of aluminum solid solution (Fig. 7.3.6).

Hardness: Average Vickers hardness at 10 kg load was 110 HV.

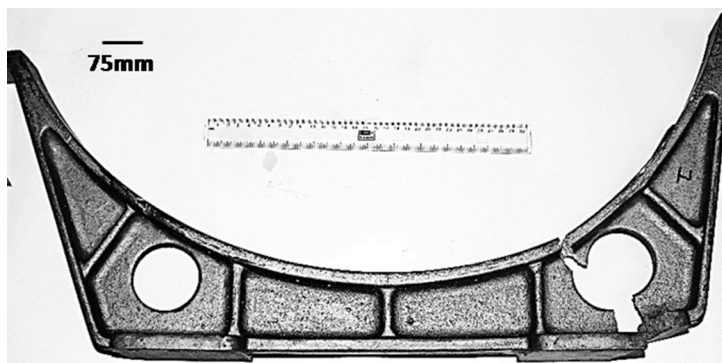
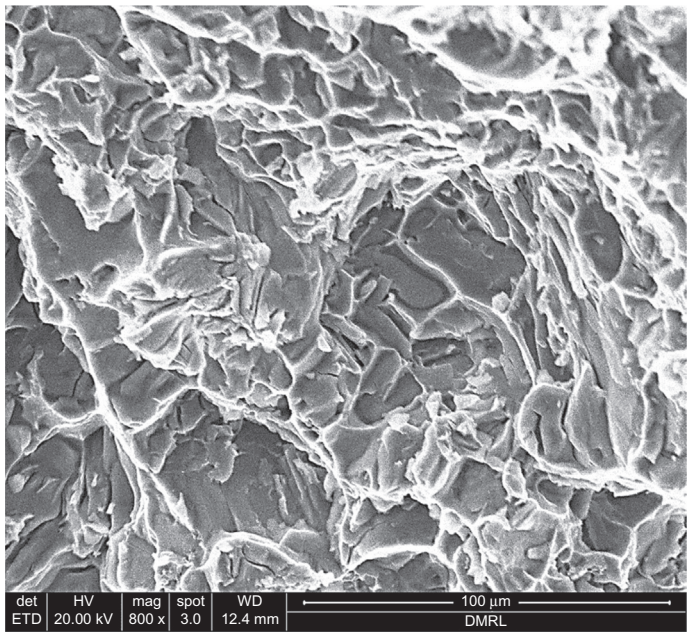


Fig. 7.3.1 Failed body of the saddle in the as-received condition.



Fig. 7.3.2 Fracture surface of the failed component showing a mixed mode of granular and fibrous fracture feature.



det	HV	mag	spot	WD	100 μm
ETD	20.00 kV	800 x	3.0	12.4 mm	DMRL

Fig. 7.3.3 SEM fractograph of the fracture surface showing region of quasi-cleavage fracture.

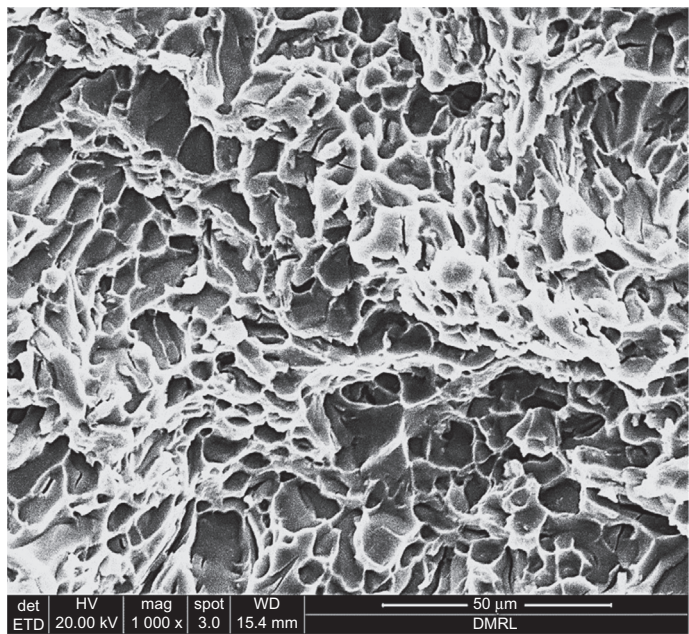


Fig. 7.3.4 SEM fractograph of the fracture surface showing region of dimpled rupture.

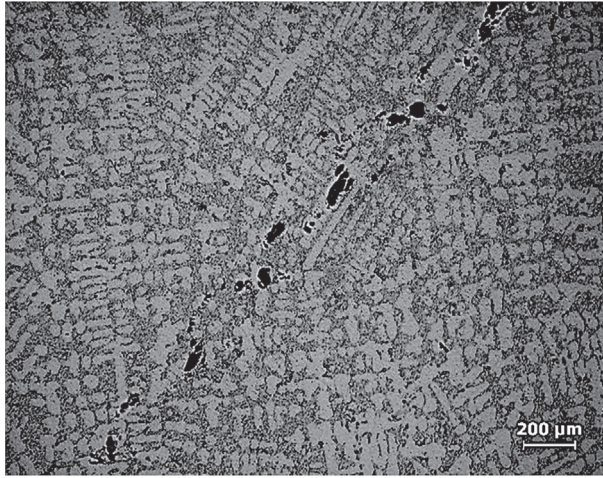
7.3.3 Chemical Analysis (Wt.%)

Si	Mg	Mn	Fe	Al
12.15	0.22	0.29	0.49	Balance

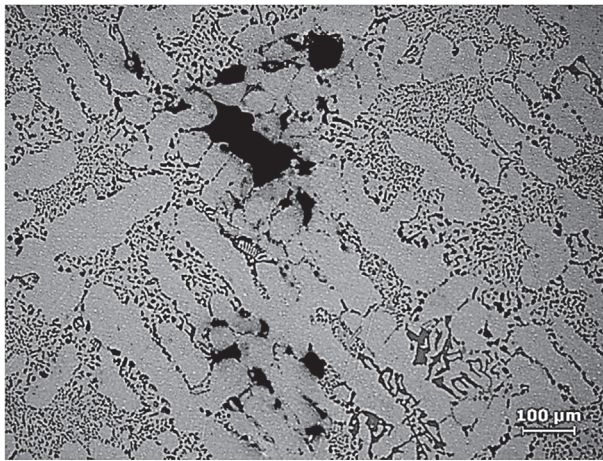
7.3.4 Discussion

The chemical composition of the failed component nearly conformed to an aluminum casting grade AK9ch (Al 4) of specification GOST-1583-93. Metallographic examination showed a number of shrinkage cavities and porosities in the casting. The microstructure and hardness suggested that the casting was in a solution treated and artificially aged condition.

Mixed mode fracture comprising of quasi-cleavage and dimples all over the fracture surface indicated that the failure had taken place due to overload. The shrinkage cavities and gas porosities present in the casting could have aided in the crack propagation.



(A)



(B)

Fig. 7.3.5 Micrograph of longitudinal and transverse sections of the casting showing (A) aligned gas porosities as a stringer and (B) as pockets. Unetched.

7.3.5 Conclusion

1. The body of the saddle assembly had failed due to overload. The casting defects could have aided in the fracture.
2. There was no gross material deficiency as far as the composition, microstructure, and hardness are concerned. However, metallurgical abnormality in the form of casting defects calls for attention and reduction in the same is expected in such castings.

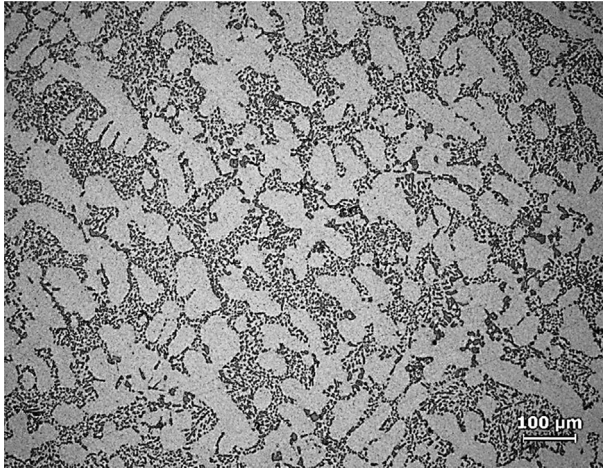


Fig. 7.3.6 Micrograph of a section of the casting showing dendrites of aluminum solid solution, some particles of Mg_2Si (black), and interdendritic network of eutectic silicon (gray). Etchant: Keller's reagent.



7.4 CASE STUDY: FAILURE OF BOILER TUBE OF A THERMAL POWER STATION

7.4.1 Background

A water wall tube of 210MW boiler of unit II of a thermal power station failed prematurely. The failure was located at the horizontal portion of 44th tube from the right-hand side rear arch tube. The tube was designed to operate at a temperature of 300–400°C and at a pressure of 160 kg/cm². The tubes were stated to be conforming to ASTM-A210.

7.4.2 Experimental Results

Visual Examination: The failure appeared to have originated on the fire side of the tube and propagated longitudinally. The fracture appeared to be a characteristic 'thin lip' rupture. The tube had bulged considerably near the failed region and the wall thickness of the tube was found to have reduced from 8mm to nearly 2.5mm. A photograph of the failed tube is presented in [Fig. 7.4.1](#).

Fractographic Examination: A 20mm longitudinal section containing the fracture surface was cut from the region showing maximum deformation. The fracture surface was ultrasonically cleaned in acetone and examined



Fig. 7.4.1 Failed water wall tube in the as-received condition.

under SEM. Slightly elongated dimples of different sizes were noticed on the fracture surface (Fig. 7.4.2).

Metallographic Examination: A tangential section containing the crack tip and fracture edges was prepared for the metallographic examination. On etching with 2% nital, the sample revealed globular carbides in a ferrite matrix (Fig. 7.4.3). A 10 mm long transverse section near the region of maximum deformation containing the fracture surface was prepared for metallographic examination. On etching with 2% nital, the sample revealed elongated grains of ferrite and globular carbides near the fracture surface

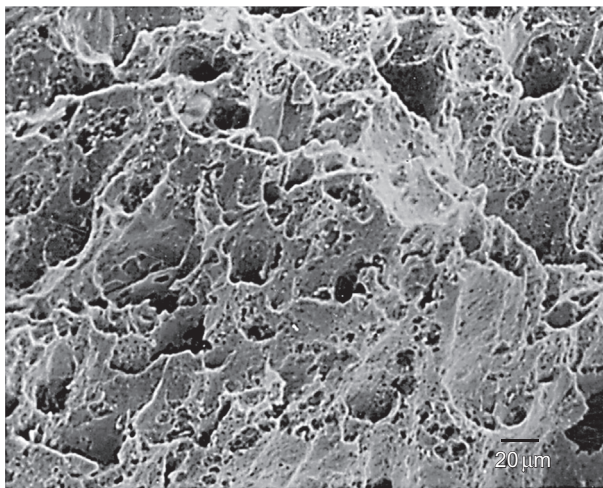


Fig. 7.4.2 SEM fractograph of the fracture surface showing a dimpled rupture.

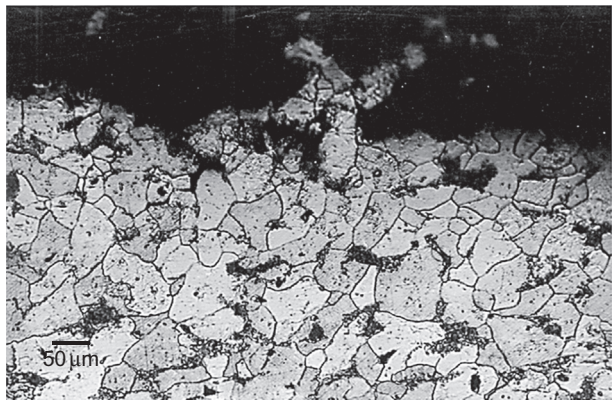


Fig. 7.4.3 Micrograph of a tangential section of the sample showing globular carbides and nearly equiaxed ferrite. Etchant: 2% nital.



Fig. 7.4.4 Micrograph of a transverse section of the sample near the region of maximum deformation showing severely deformed ferrite and globular carbides. Etchant: 2% nital.

(Fig. 7.4.4). Metallographic examination of a transverse section taken from the rear side of the tube revealed equiaxed ferrite and lamellar pearlite structure (Fig. 7.4.5).

Hardness: Hardness was measured at 2.5 kg load on the samples used for metallographic examination and the results are given here.

	Sample	Hardness (HV)
(i)	Near the fracture surface	173
(ii)	Away from the failed region	129

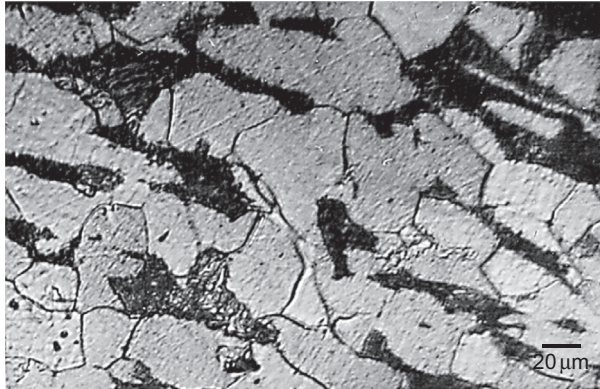


Fig. 7.4.5 Micrograph of a transverse section of the sample from rear side of the tube showing equiaxed ferrite and lamellar pearlite. Etchant: 2% nital.

7.4.3 Chemical Analysis (Wt.%)

C	Si	Mn	S	P	Fe
0.21	0.27	0.58	0.015	0.018	Rem.

7.4.4 Discussion

Chemical composition did not conform to the specified steel as the carbon content was much lower than that was needed. Visual examination of the failed water wall tube showed considerable thinning along its length on the fire side and bulging near the failed region. These indicated that the tube had undergone considerable plastic deformation prior to failure. Presence of elongated ferrite grains near the region of maximum deformation indicated that the deformation had taken place below the recrystallization temperature of ferrite. Presence of globular carbides near the fracture region and lamellar pearlite at regions diametrically opposite to the failed region indicated that the fire side of the tube was subjected to prolonged overheating. The presence of dimples on the fracture surface indicated that the failure had occurred due to overload.

The hoop stress and radial stress calculated at the inner wall of the tube at an operating pressure of 160 kg/cm^2 were 7.05 kg/mm^2 and 1.61 kg/mm^2 , respectively. The yield strength of the steel comes down drastically with increase in operating temperature. Application of yield criterion indicated that the operating stress exceed the yield strength of the material at about

620°C. It is reasonable to conclude that the tube was operated at such a high temperature as indicated by the changes in microstructure.

7.4.5 Conclusion

The failure of the tube was due to overload as it was operated at temperatures above the designed operating temperature, due possibly to localized overheating.



Fatigue

Parts that are designed for rotating, vibrating, oscillating, bending, moving, or flexing are subjected to cyclic loading conditions. Failures occurring under conditions of cyclic or dynamic loading are called fatigue failures. Aircraft engine components viz. compressors, turbines, gears, shafts, bearings and pumps, and automobile parts are some of the very large mechanical devices that are subjected to repeated loading and vibrations. Fatigue is stated to account for about 75% of all service failures due to mechanical cause.

Unlike failures under static loads fatigue failures occur at lower stresses that are lower than the yield strength or ultimate tensile strength of the material. Fatigue failures occur mainly in three stages. Accumulation of defects and nucleation of micro cracks called the initiation zone, propagation of fatigue crack called the propagation zone, and the final failure zone where the remaining cross section of the part is no longer able to withstand the applied load and hence fails due to overload.

Fatigue failures are generally found to start on the surface where there is a local stress concentration. A stress concentration can occur at (a) sharp changes in the part such as fillets, holes, grooves, keyways, etc., (b) rough surface, tool marks, machining marks, etc., and (c) internal defects in the material such as voids, porosities, blow holes, nonmetallic inclusions, decarburized layer, etc.

Fatigue is a function of the number of stress cycles and the amplitude of stress. At lower levels of stress, the metal will be able to withstand a larger number of stress cycles whereas at higher levels of stress the metal will fail at relatively lower number of stress cycles. Endurance limit is the level of stress below which fatigue does not occur regardless of the number of stress cycles. Many metals, particularly steels, have endurance limits. Most of the nonferrous metals do not have an endurance limit. Hence the fatigue life of such metals is determined as the stress level that would withstand 10^7 or 10^8 stress cycles. The S-N curve is the popular method to estimate the fatigue strength of a metal.

A fatigue failure occurs without any prior warning and therefore it calls for a lot more caution than any other type of failure. Failure results in a brittle fracture without any gross deformation near the fracture. The fracture surface leaves certain telltale evidence in case of a fatigue failure. The fracture

surface could show a smooth region, due to the rubbing action of the mating fracture surfaces during crack propagation, and a fibrous region which is indicative of final failure occurring under ductile overload condition. Frequently the progress of fracture is seen as 'beach marks' that are a series of rings of progressing size from the point of initiation. Fatigue fracture and fractography of fatigue fracture are described in some detail in Sections 2.3 and 2.4. The reader is advised to read more about fatigue as mentioned in bibliography at the end of the book.



8.1 CASE STUDY: FAILURE OF WHEEL HUB OF AN AIRCRAFT

8.1.1 Background

A piece of the wheel hub of a main wheel assembly of an aircraft broke and fell on the runway during take-off. The wheel hub had completed 591 landings against a total life of 1500 landings. It was also gathered that the component had completed a life of 938 h against a total overhaul life of 1500 flying hours. The main wheel hub was fitted with a tubeless tire.

8.1.2 Experimental Results

Visual Examination: The wheel hub in the as-received condition is shown in Fig. 8.1.1. Careful examination showed that the fracture was initiated at the junction of the rim and hub on the outer surface of the wheel. On examination under stereomicroscope, the fracture surface showed

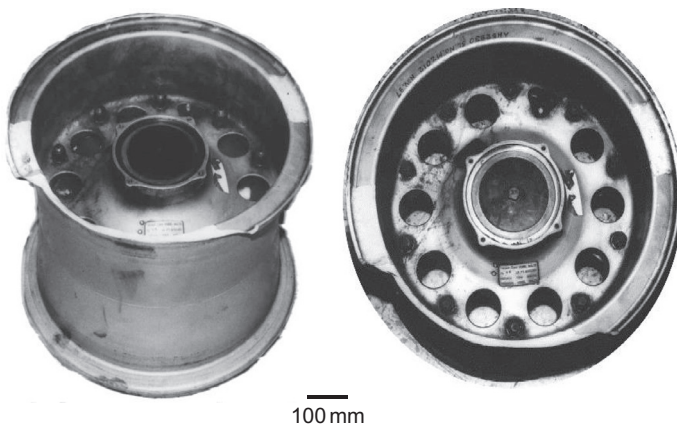


Fig. 8.1.1 Failed wheel hub in the as-received condition (two views).

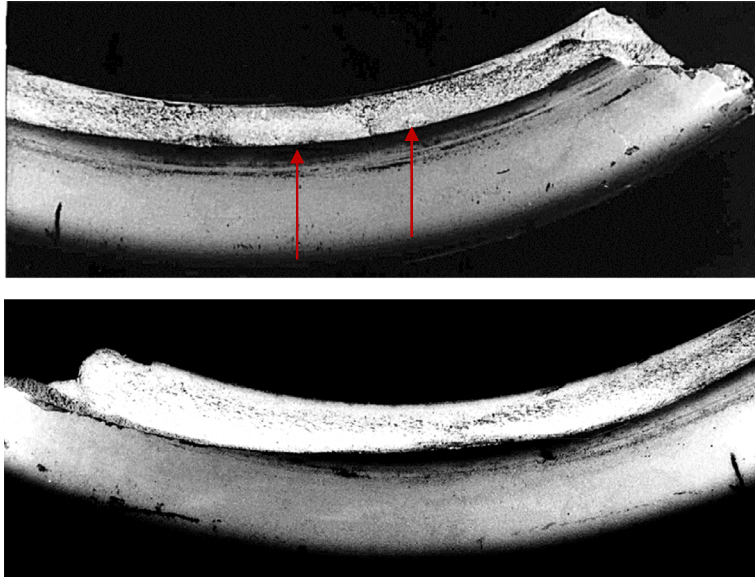


Fig. 8.1.2 Close-up view of the fracture surface showing two origins 30mm apart (arrow heads).

two origins of fracture, which were about 30 mm apart (Fig. 8.1.2, arrow head). Fracture surface showed smooth features to a depth of about three-fourth of the section starting from the origin. The remaining fracture was fibrous.

Fractographic Examination: On examination of the cleaned fracture surface under scanning electron microscope (SEM), small pockets of inclusion clusters were noticed at the origin (Fig. 8.1.3). Beach marks originated from these pockets of inclusions. On analyzing the inclusion cluster by EDAX, it was found to contain oxygen and chlorine besides aluminum and copper (Fig. 8.1.4). Characteristic fatigue striations formed the essential feature of the smooth fracture region when viewed at higher magnification (Fig. 8.1.5). The remaining fracture revealed quasi-cleavage fracture features (Fig. 8.1.6).

Metallographic Examination: Sections of the wheel hub were prepared for metallographic examination. On etching with Keller's reagent elongated grains of aluminum solid solution with undissolved precipitates were revealed (Fig. 8.1.7).

Hardness: At 10 kg load the average Vickers hardness of the failed component was 150HV.

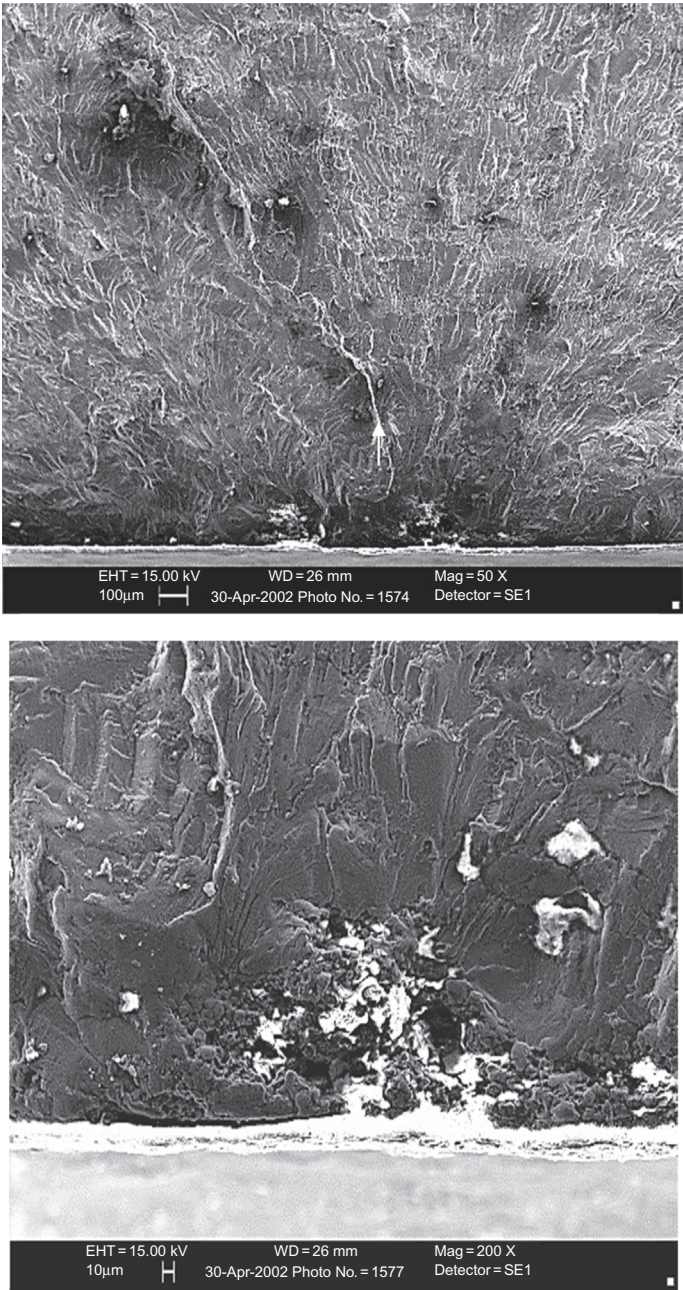


Fig. 8.1.3 SEM fractographs showing cluster of inclusions at the origin.

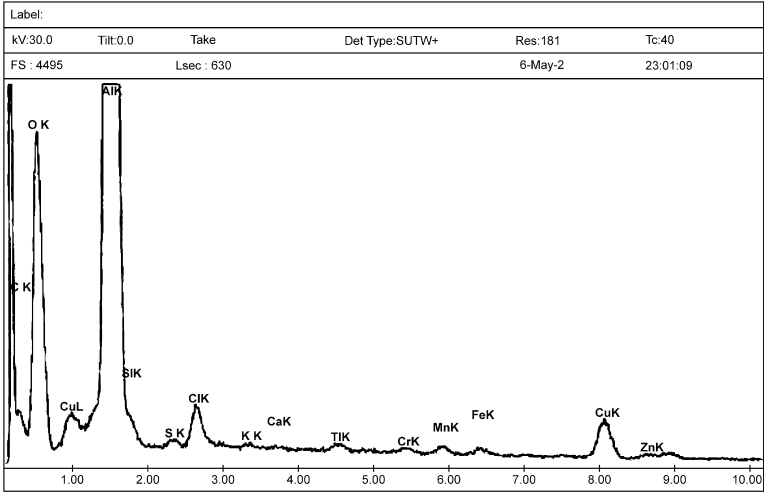


Fig. 8.1.4 EDAX of the inclusion at the origin.

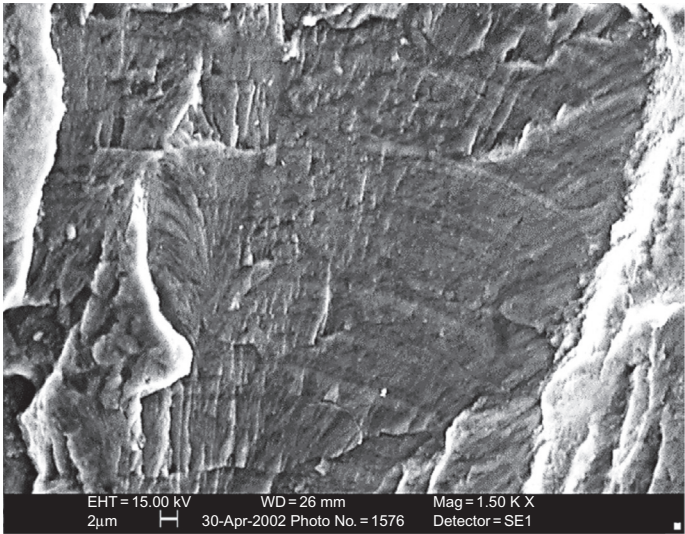


Fig. 8.1.5 SEM fractograph of the smooth fracture region showing fatigue striations.

8.1.3 Chemical Analysis (Wt.%)

Cu	Mn	Mg	Si	Al
3.50	0.48	1.00	0.73	Rem.

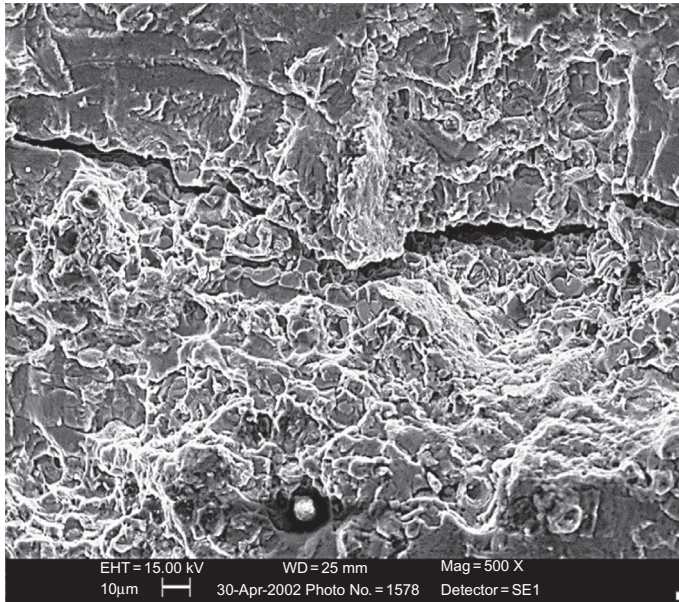


Fig. 8.1.6 SEM fractograph of the remaining fracture surface showing quasi-cleavage fracture.

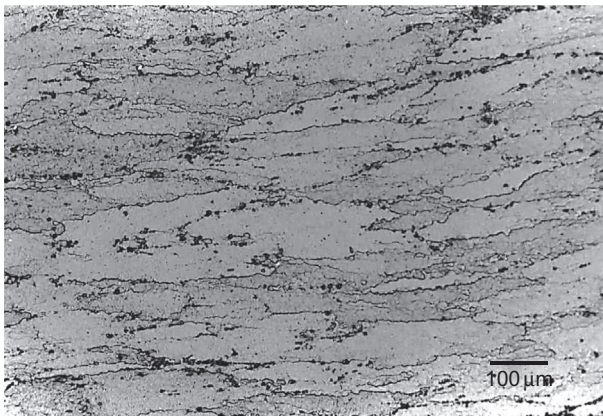


Fig. 8.1.7 Micrograph of the wheel hub sample showing elongated grains of aluminum solid solution with undissolved precipitates. Etchant: Keller's reagent.

8.1.4 Discussion

Chemical composition, metallography and hardness suggested that the component was forged from an Al-Cu-Mg-Mn alloy nearly equivalent to BS 2L 77, 2014A and used in a solution treated and aged condition. The analysis of

the inclusion cluster at the fracture origin indicated that it was a slag entrapment in the alloy. The slag inclusion caused the initiation of fatigue crack that propagated under cyclic loading condition, which remained concealed under the tire and was not detected during the previous stages of examination. The final failure was due to overload.

8.1.5 Conclusion

The wheel hub had failed due to fatigue initiated at two slag pockets entrapped at the junction of the rim and hub on the outer surface of the wheel.



8.2 CASE STUDY: FAILURE OF BOLTS OF FLAPJACK MOUNTING BRACKET OF AN AIRCRAFT

8.2.1 Background

Two bolts used to retain the flapjack-mounting bracket were found broken and lodged in the access panel of the flapjack. Though small components these bear significance, from the point of view of their life, as it was considered a premature failure.

8.2.2 Experimental Results

Visual Examination: On cleaning rust and grease from the fracture surfaces they revealed beach marks (Fig. 8.2.1).

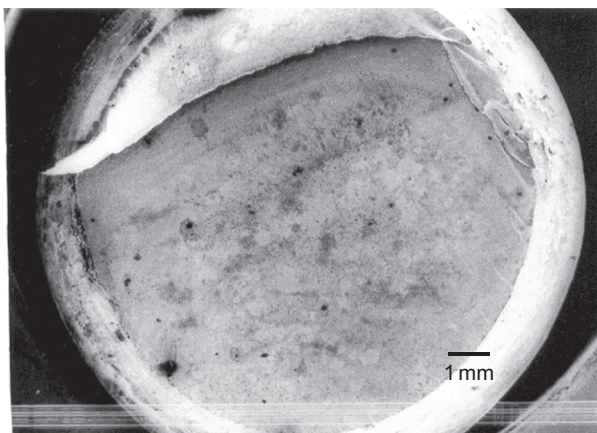


Fig. 8.2.1 Fracture surface of one of the bolts showing beach marks.

Fractographic Examination: When examined under SEM the fracture was found to have originated from the thread root and propagated across the section. The smooth fracture occupying two-third of the fracture section revealed beach marks at lower magnification and fatigue striations at higher magnification (Fig. 8.2.2). The remaining fracture surface revealed equiaxed dimples (Fig. 8.2.3).

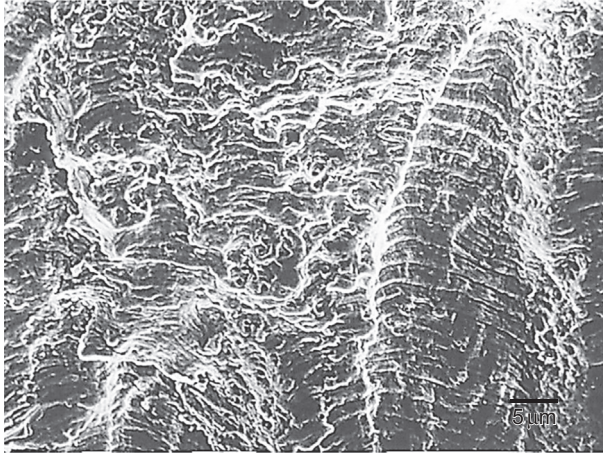


Fig. 8.2.2 SEM fractograph of the region containing beach marks showing fatigue striations at higher magnification.

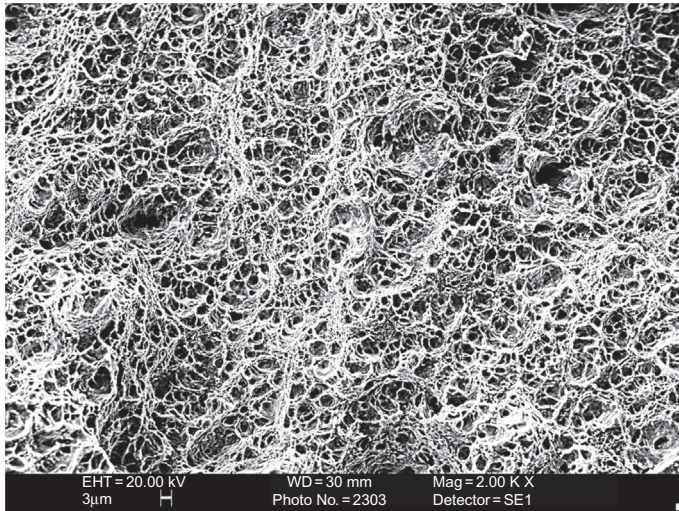


Fig. 8.2.3 SEM fractograph of the final fracture region showing equiaxed dimples.

Metallographic Examination: Longitudinal section of the bolt samples showed decarburized layer to a depth of 0.15 mm all over below the threads (Fig. 8.2.4). The microstructure revealed tempered martensite (Fig. 8.2.5).

Hardness: Average micro hardness measured at 100 g load at decarburized layer was 280 HV, whereas the average bulk hardness of the bolt measured at 10 kg load was 375 HV.

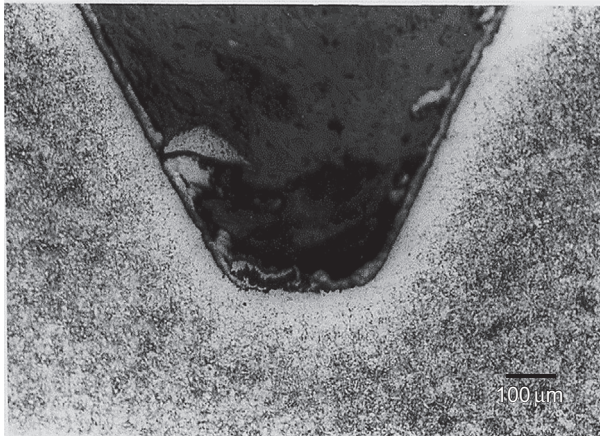


Fig. 8.2.4 Micrograph of the bolt sample showing decarburized layer all-along the threaded region. Etchant: 2% nital.

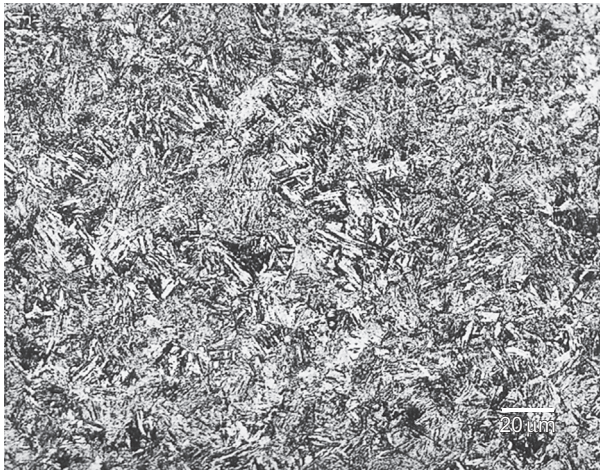


Fig. 8.2.5 Micrograph of a section of the bolt sample showing a tempered martensite structure. Etchant: 2% nital.

8.2.3 Chemical Analysis (Wt.%)

C	Si	Mn	Cr	S	P	Fe
0.37	1.26	0.87	0.93	0.012	0.02	Rem.

8.2.4 Discussion

Chemical analysis suggested that the bolts were made from an Si-Mn-Cr containing low alloy steel nearly conforming to 30XГCA of specification GOST 4543-71(Russian). The microstructure and hardness indicated that the bolts were used in a hardened and tempered condition. A decarburized layer of 150 μm below the thread points to the fact that the bolts were hardened after the threads were cut on them resulting in their exposure to the furnace atmosphere during austenitization. Lower hardness of the decarburized layer was the source of fatigue crack initiation and its propagation under cyclic loading conditions. The final failure of the bolt was due to overload, as the remaining cross-sectional area could no longer withstand the applied load.

8.2.5 Conclusion

The bolts of the flapjack-mounting bracket had failed due to fatigue initiating at the decarburized layer below the thread root.



8.3 CASE STUDY: FAILURE OF COMPRESSOR ROTOR BLADES OF AN AEROENGINE

8.3.1 Background

Aeroengine of a helicopter was received at a base repair depot with a defect of heavy abnormal background noise while starting the aeroengine. During strip examination of engine compressor rotor blades of fourth stage were found damaged. No other information viz. material grades, life completed before failure, total technical life, etc., were given by the depot.

8.3.2 Experimental Results

Visual Examination: The failed fourth stage compressor rotor blade in the as-received condition is shown in [Fig. 8.3.1](#). Convex surface of the aerofoil region showed dark brown deposits near the leading edge and a few pits.

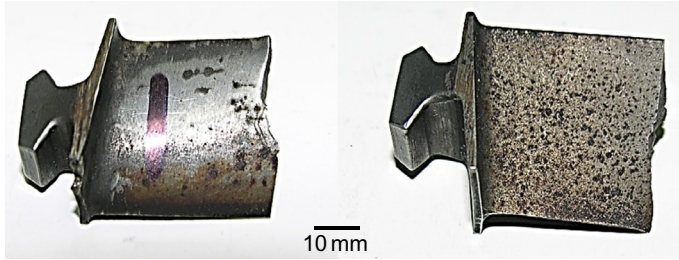


Fig. 8.3.1 Compressor rotor blade in the as-received condition showing the convex and the concave surface of the aerofoil containing a large number of corrosion pits and brown deposits near the leading edge.

Concave surface of the aerofoil region showed a large number of pits and brown deposits. The fracture surface showed smooth fracture containing beach marks to a length of two-third from the leading edge. The final one-third of the fracture surface showed fibrous fracture features.

Fractographic Examination: The fracture surface was cleaned ultrasonically in acetone and examined under scanning electron microscope (SEM). The fracture surface showed corrosion pit at the leading edge and smooth fracture with beach marks (Figs. 8.3.2 and 8.3.3). The debris within the corrosion pit was analyzed by energy dispersive X-ray analysis (EDAX). Presence of chlorine, sulfur, and bromine was found in the pitted region besides

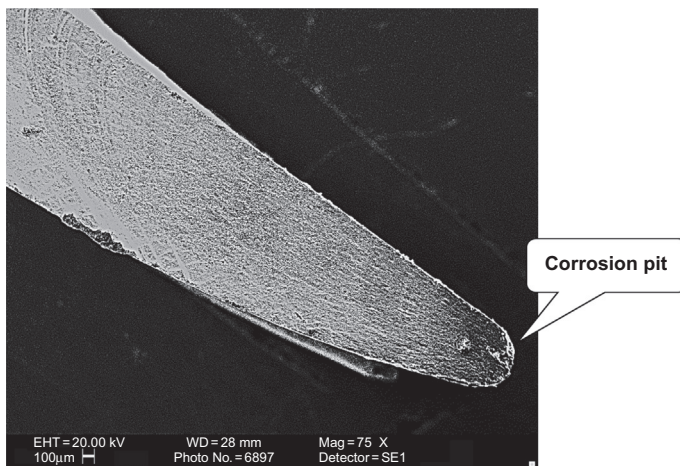


Fig. 8.3.2 SEM fractograph of the fracture surface showing corrosion pit at the leading edge and smooth fracture surface with beach marks.

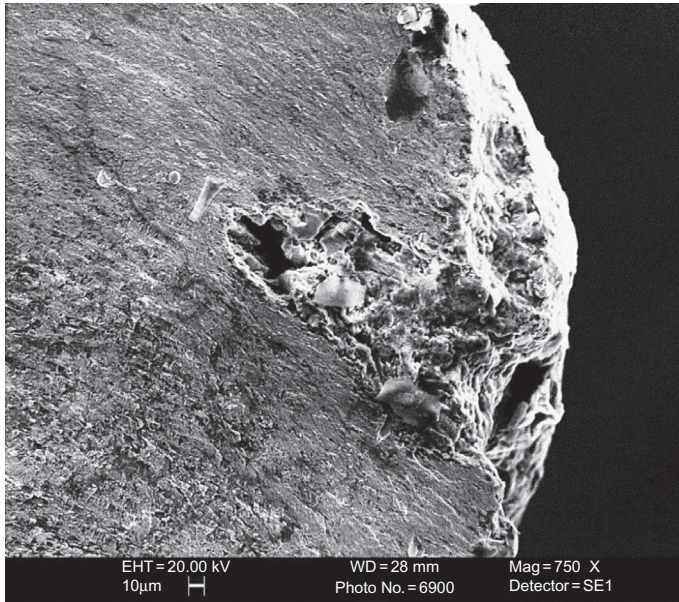


Fig. 8.3.3 SEM fractograph of the fracture surface at higher magnification showing corrosion pit at the leading edge.

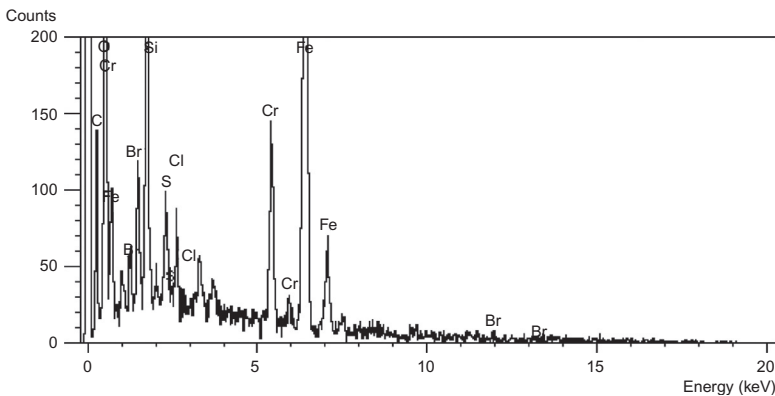


Fig. 8.3.4 EDAX spectrum of the debris within the corrosion pit at the leading edge showing the presence of sulfur, chlorine, and bromine besides the alloying elements of the blade.

chromium and iron (Fig. 8.3.4). From near the pit beach marks were found initiating and propagating toward the trailing edge (Fig. 8.3.5). Fatigue striations were revealed in the area showing beach marks (Fig. 8.3.6). The final fracture region revealed equiaxed dimples (Fig. 8.3.7).

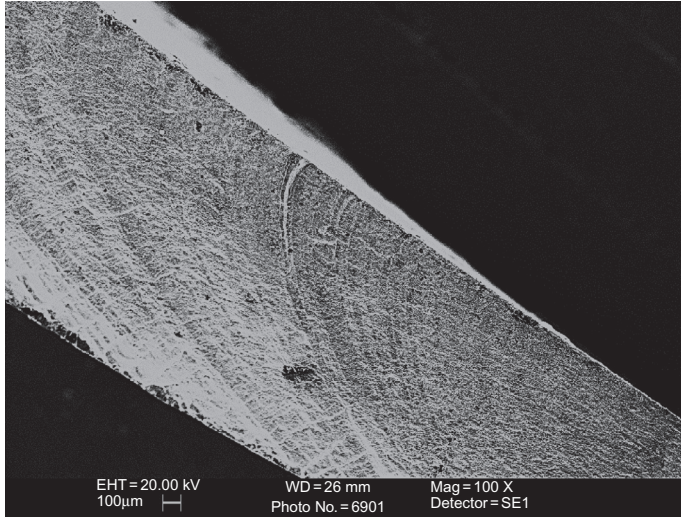


Fig. 8.3.5 SEM fractograph of the fracture surface showing beach marks initiating near the corrosion pit at the leading edge and propagating toward the trailing edge.

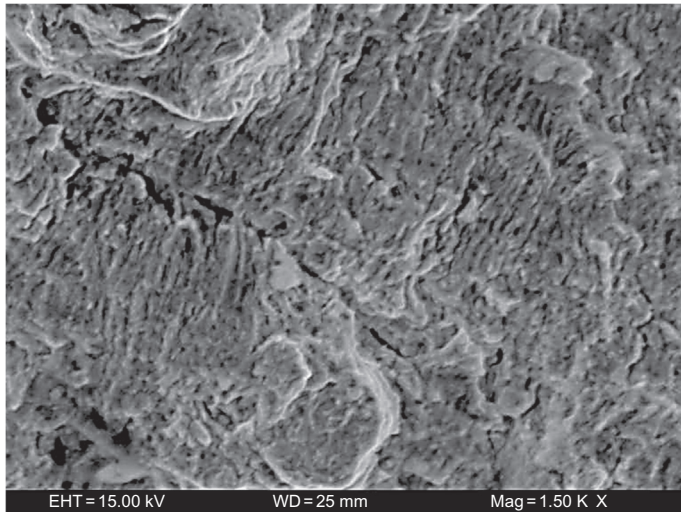


Fig. 8.3.6 SEM fractograph of the fracture surface showing fatigue striations.

Metallographic Examination: Section across the aerofoil parallel to the fracture surface was prepared for metallographic examination. In the unetched condition, the sample showed a number of pits on the concave surface of the aerofoil near the leading edge and a few pits on the convex surface of the

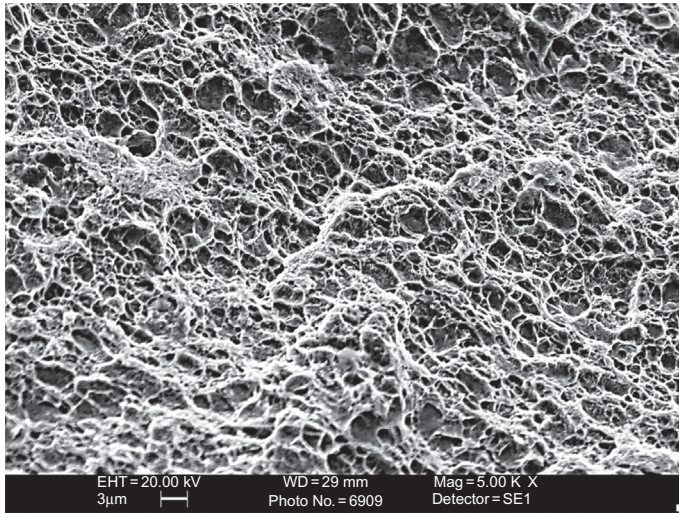


Fig. 8.3.7 SEM fractograph of the fracture surface showing equiaxed dimples at the final fracture zone.

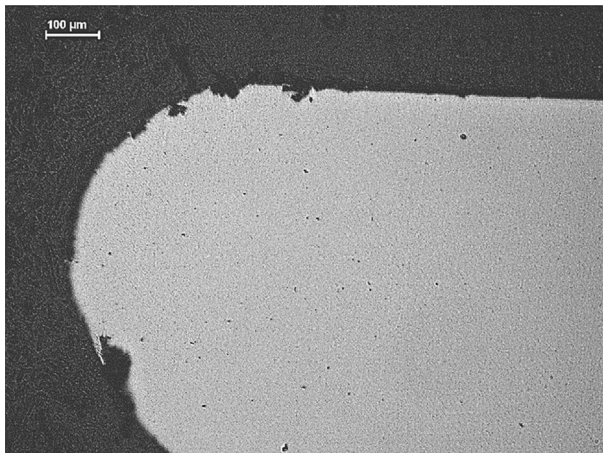


Fig. 8.3.8 Micrograph of transverse section of the aerofoil showing pitting corrosion near the leading edge. Unetched.

aerofoil near the trailing edge (Fig. 8.3.8). On etching with aqua regia, the sample revealed a tempered martensite structure (Fig. 8.3.9).

Hardness: Average Vickers hardness of the sample measured at 20 kg load was 376 HV.

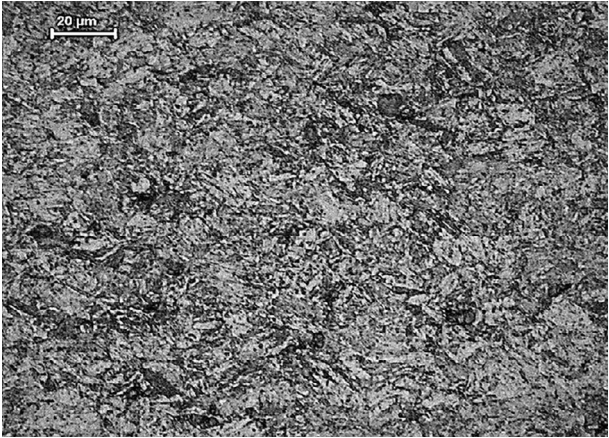


Fig. 8.3.9 Micrograph of transverse section of the aerofoil showing a tempered martensite structure. Etchant: Aqua regia.

8.3.3 Chemical Analysis (Wt.%)

C	Si	Mn	Ni	Cr	S	Fe
0.145	0.50	0.60	2.24	10.63	0.01	Balance

8.3.4 Discussion

The failed compressor rotor blade was made of martensitic stainless steel that did not conform to any standard grade and used in a hardened and tempered condition. The blade had suffered severe pitting corrosion as evidenced by pits and brown deposits seen on the aerofoil. Pitting corrosion was confirmed by the EDAX of the debris within the pit, which revealed the presence of corrosive substances like sulfur, bromine, and chlorine. Large pit near the aerofoil acted as site for the initiation of fatigue crack that propagated under fatigue loading conditions during service to a length of three-fourth of the fracture surface as evidenced by SEM fractography. The remaining section over one-third of the length could no longer withstand the applied load and failed due to overload as evidenced by dimpled rupture.

8.3.5 Conclusion

The fourth-stage compressor rotor blade had failed due to fatigue that was initiated at corrosion pits at the leading edge.



8.4 CASE STUDY: FAILURE OF BEARING CAGE OF AN AEROENGINE

8.4.1 Background

Aeroengine of a helicopter was received at a base repair depot for defect investigation. It was reported that the subject aeroengine produced heavy abnormal background noise while starting. During strip examination of the engine the first support bearing cage of ball bearing was found damaged.

The users have not given any information regarding completed life and total technical life, material specification and manufacturing details, etc. of the failed component.

The other parts of the ball bearing viz. the inner race, the outer race, and the balls were found intact and hence were not taken up for any tests.

8.4.2 Experimental Results

Visual Examination: The failed ball bearing showing the inner race, the outer race, the balls, and the cage in the as-received condition are shown in Fig. 8.4.1. The cage showed breakage at two locations: one parallel and the other inclined at 45° to the axis of shaft (Fig. 8.4.2). The fracture surfaces at the straight crack showed smooth fracture, while the crack inclined at 45° showed fibrous fracture.

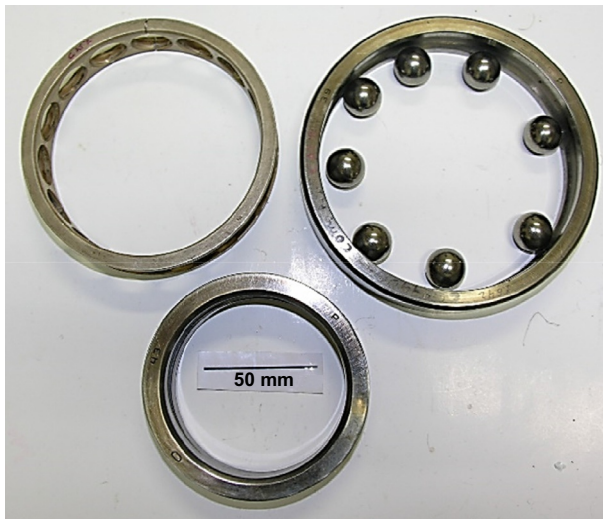


Fig. 8.4.1 The failed ball bearing showing the inner race, the outer race, the balls, and the cage in the as-received condition.

The circular slots where the balls are seated within the cage showed the path traced by the balls while in motion. This path was visibly off-center in all the circular slots (Fig. 8.4.3). Moreover, the path also showed two dent marks at 3 and 9'O clock positions with respect to shaft axis (Fig. 8.4.4).

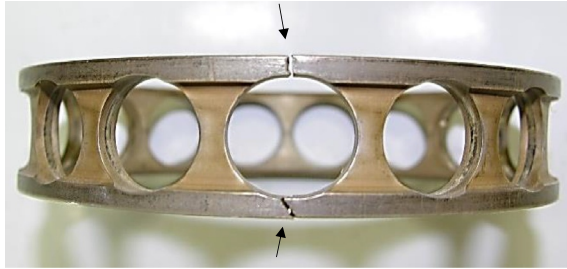


Fig. 8.4.2 The broken cage in the as-received condition showing one of the breakage parallel and the other inclined at 45° to the axis of shaft.



Fig. 8.4.3 The broken cage in the as-received condition showing the eroded track marks of the ball within each circular slot and the dent marks at two opposite locations.



Fig. 8.4.4 Magnified image of the track mark and the dent mark with pointed ends.

The smooth crack is observed to have initiated at the dent marks from inner surface and propagated to the outer surface of the circular slot.

Fractographic Examination: The fracture surfaces of both straight fracture and inclined fracture at 45° were cleaned ultrasonically in acetone and examined under scanning electron microscope (SEM). The fracture surface of the straight fracture showed smooth brittle fracture over three-fourth of the section and fibrous fracture over the remaining area (Fig. 8.4.5). The crack is observed to have originated at the inner surface and propagated toward the outer surface (Fig. 8.4.6). The smooth fracture region revealed ill-defined striations at higher magnification (Fig. 8.4.7). The remaining one-fourth region on the fracture surface revealed elongated dimples (Fig. 8.4.8). The fracture surface of 45° inclined fracture revealed predominantly dimpled rupture (Fig. 8.4.9).

Metallographic Examination: A section of the cage across the dent marks was prepared for metallographic examination. In the unetched condition, the sample showed surface erosion and pitting on the inner surface (Fig. 8.4.10A). The outer surface was relatively intact (Fig. 8.4.10B). On etching the sample with acidified potassium dichromate solution, the sample

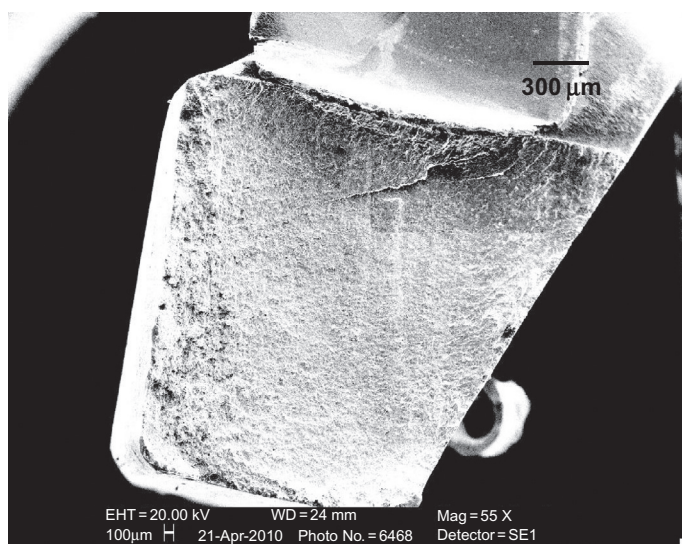


Fig. 8.4.5 SEM fractograph of the straight fracture surface showing brittle smooth fracture over three-fourth of the region and fibrous fracture over the remaining region.

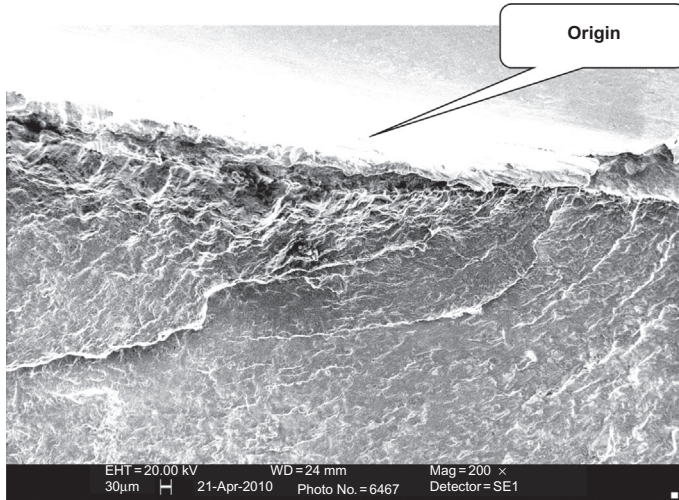


Fig. 8.4.6 SEM fractograph of the fracture surface of the cage showing crack origin and beach marks diverging from the origin.

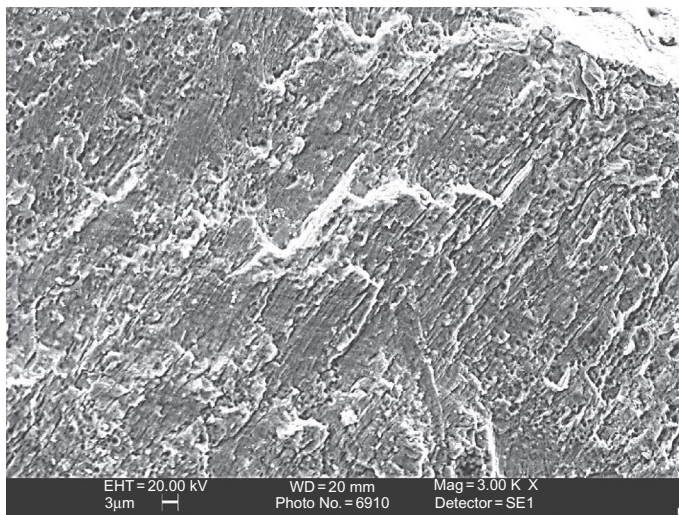


Fig. 8.4.7 SEM fractograph of the smooth region of the fracture surface revealing fatigue striations.

revealed a cast structure having alpha grains in a matrix of eutectoid decomposed beta phase and retained beta phase (Fig. 8.4.11).

Hardness: Average Vickers hardness of the cage sample measured at 5 kg load was 199 HV.

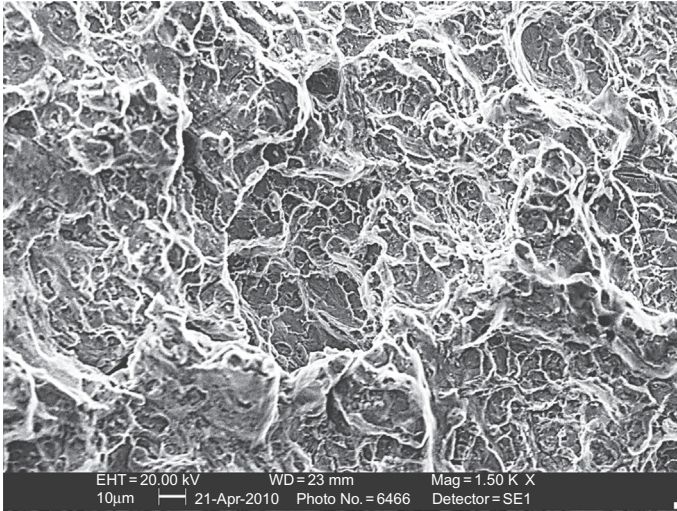


Fig. 8.4.8 SEM fractograph of the fibrous region of the straight fracture surface revealing dimples.

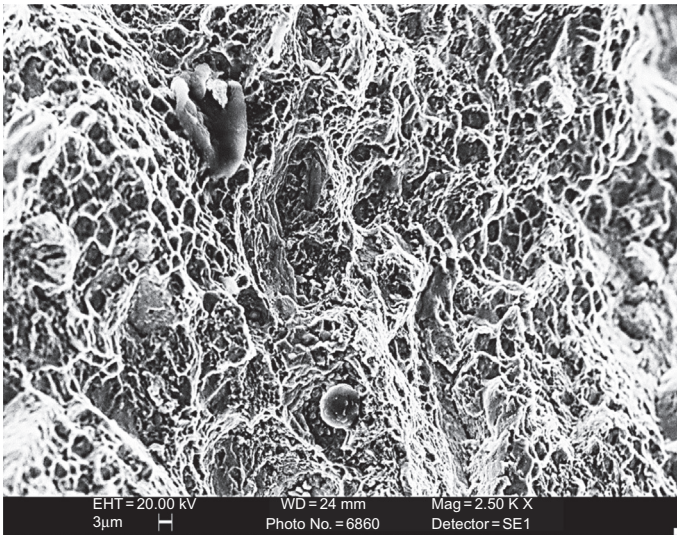


Fig. 8.4.9 SEM fractograph of the fracture surface inclined at 45° revealing, essentially equiaxed dimples.

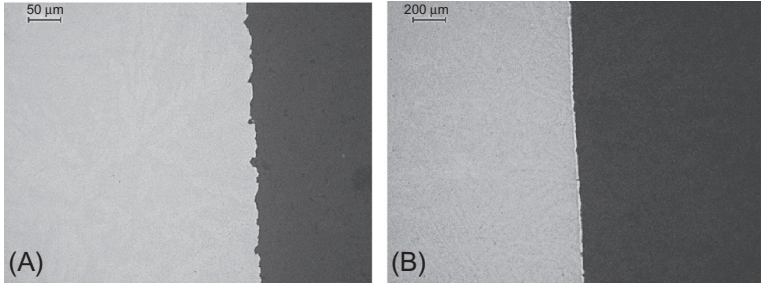


Fig. 8.4.10 Micrograph of a section of the cage showing (A) erosion and pitting on the inner surface and (B) relatively intact outer surface. Unetched.

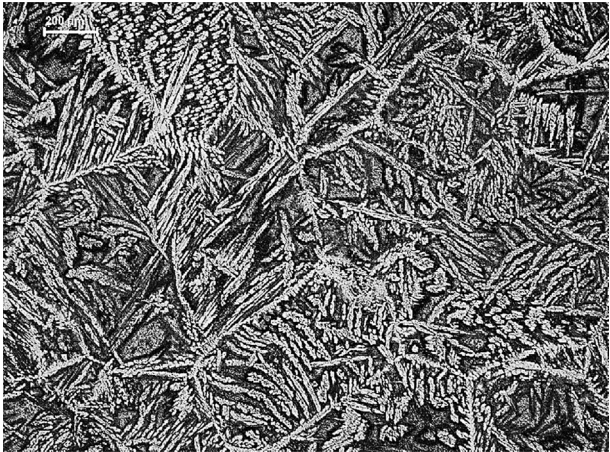


Fig. 8.4.11 Micrograph of a section of the cage showing cast structure having alpha grains in a matrix of eutectoid decomposed beta phase and retained beta phase. Etchant: Acidified potassium dichromate.

8.4.3 Chemical Analysis (Wt.%)

Al	Ni	Fe	Cu
9.73	5.05	4.0	Balance

8.4.4 Discussion

Chemical composition suggested that the failed cage was a casting of aluminum bronze conforming to specification BS 1400: 1985, grade AB-2. Visual examination showed that the path traced by the balls were off-center within the circular slots in the cage. This indicated that during service the balls were pushed to one side of the center. Moreover, the dent marks on the 3 and 9 o'clock positions with respect to the shaft axis on the ball path indicated that the balls were subjected to localized axial loading while in motion. The dent mark revealed sharp edges as evidenced in stereo microscopic examination. Metallographic examination confirmed erosion and pitting of the inner surface of the circular slot.

Local erosion of the dented portion on the ball path initiated fatigue crack that propagated under compressive fatigue loading condition, to three-fourth of the cage section. The remaining one-fourth section failed due to overload. Fractographic examination confirmed that the failure at the straight cracked region was due to fatigue. Fatigue cracking of the cage at one location caused overload on the weak zone of opposite side resulting in overload failure of this location as evidenced by 45° shear and dimpled rupture.

8.4.5 Conclusion

The ball-bearing cage had failed due to fatigue which had initiated at the dent mark on the ball path as a result of surface erosion and localized axial load. The fatigue failure on one location caused overload failure of the cage on the opposite location.



Corrosion

Corrosion is defined as the deterioration of a metal due to chemical or electrochemical reactions with its environment. The cost of corrosion and of corrosion prevention is approximately 5.0% of the gross national product of any country, hence is of enormous economic cost to society.

There are different types of corrosion, which a metal can be affected with: they are—uniform corrosion, pitting corrosion, crevice corrosion, intergranular corrosion, hot corrosion, and stress corrosion cracking.

(A) *Uniform corrosion*: It is the most common variety of corrosion such as rusting of iron. Other metals and alloys also corrode uniformly such as aluminum, copper, brass, and magnesium. However, depending upon the nature of the metal and the environment, some of them will form either protective or nonprotective films on the surface. Since uniform corrosion of different metals and alloys has been studied at length, it is fairly predictable and controlling the same is relatively simple. Use of more corrosion-resistant metal for a given environment and adopting suitable surface protection measures such as painting, coating, anodizing, plating, cladding, and cathodic protection are commonly used for preventing a metal from uniform corrosion.

(B) *Pitting Corrosion*: It is a form of extremely localized attack that results in the formation of a pit leading to perforation in the metal. It is one of the insidious forms of corrosion as it results in failure of a part without much of a weight loss. Pitting takes a longer period of initiation before they are visible. Once initiated, a pit, however, penetrates into the section at an ever-increasing rate.

Stagnant condition of a liquid in a tank or lower part of a pipe can initiate pitting. Most pitting is associated with halide ions, with chlorides, bromides, and hypochlorites being the most prevalent. Oxidizing metal ions with chlorides can cause faster pitting.

Selection of a more resistant material to pitting corrosion is the most important method of prevention of this problem. For example, type 316 stainless steel containing 2–3 wt.% Mo is more resistant to pitting as

compared to type 304 stainless steel. As pitting is associated with local break down of passive films that protect the metal it is needless to say that any damage caused to the passive film has to be prevented in order to improve pitting resistance of the material.

(C) *Crevice Corrosion*: When metal surfaces are exposed to corrosive environment, localized corrosion takes place within crevices or shielded areas. Small volume of stagnant solution getting trapped under gaskets, lap joints, and crevices under bolt and rivet heads cause this type of attack known as crevice corrosion. To function as a corrosion site, a crevice must be wide enough to permit liquid entry, but sufficiently narrow to maintain a stagnant zone. Metals which form a passive oxide film on their surfaces for corrosion resistance are particularly prone to crevice corrosion, for example stainless steels and aluminum alloys. Chloride ions in the corrosive medium can intensify the attack. Dissimilar metals in contact can aggravate crevice corrosion attack because galvanic corrosion may also add to the problem.

There is no single method of preventing crevice corrosion, which differs with individual case and the prevailing conditions. Nevertheless, the following preventive measures could prove to be beneficial:

- (1) Use welded butt joints instead of riveted or butted joints.
- (2) Close crevices in existing lap joints by welding or soldering.
- (3) Design vessels with drain parts and avoid stagnant regions.
- (4) Remove deposited dirt, sand, oil, etc. regularly from equipment by following proper maintenance schedule.
- (5) Periodically monitor equipment during long shut downs for stagnant liquids, wet packing materials, deposits, etc. and remove them.
- (6) Use solid nonabsorbent gaskets wherever possible.

(D) *Intergranular corrosion*: This phenomenon is mainly found in stainless steels. Austenitic stainless steels (18-8 type) when heated or slowly cooled through the temperature range 400–900°C becomes sensitized. The phenomenon is called ‘Weld decay’ when the heat-affected zone near the weld in stainless steel gets sensitized. Within the forbidden temperature range of 400–900°C carbon diffuses to the grain boundaries where it preferentially combines with chromium to form chromium carbide, Cr_{23}C_6 . This causes depletion of chromium content in the areas adjoining the grain boundaries to levels much lower than that is required for corrosion resistance. When a sensitized stainless steel is exposed to corrosive environment the regions in the vicinity of the grain boundary get preferentially attacked causing an intergranular corrosion, which reduces the cohesive strength of the material and ultimately fails.

Since the phenomenon of sensitization and intergranular corrosion is very widely studied the preventive methods are also laid down clearly. The three commonly used methods to control or prevent intergranular corrosion of austenite stainless steel are as given here:

- (i) **Solution treatment:** Austenitic stainless steel should be solution heat treated at 1050–1100°C followed by rapid quenching. Soaking at high temperature allows dissolution of carbides and rapid quenching through sensitization zone prevents the precipitation of carbides.
- (ii) **Reduction of carbon content:** Austenitic stainless steels can be produced with a low carbon content generally $< 0.03\%$. Letter 'L' designates these alloys for example, 304L. Low carbon alloys have lower susceptibility to sensitization when heated or welded in the sensitizing temperature range.
- (iii) **Stabilizing the steels:** Elements like niobium and titanium have higher affinity to combine with carbon than chromium to form carbides. When added in sufficient quantity they combine with all the carbon and thus prevent formation of chromium carbide along the grain boundaries. Stainless steels thus stabilized with niobium or titanium are widely used where extensive welding is involved in order to avoid weld decay.



9.1 CASE STUDY: FAILURE OF NIPPLE OF FUEL PIPE LINE

9.1.1 Background

Nipple of a fuel pipeline connecting a 'T' joint to the fuel solenoid valve inlet to one of the reaction control thrusters developed a crack and consequent leak when it was subjected to pneumatic pressure of 1 kg/cm^2 . The users reported that the specified material for making the nipple was AISI 304 grade austenitic stainless steel in annealed condition. It was also gathered from the users that the nipple was subjected to hydraulic pressure test at stipulated levels prior to the pneumatic test that was carried out after certain time period.

9.1.2 Experimental Results

Visual Examination: A fine circumferential crack of about 5 mm length was observed at about 8 mm from the weld region (Fig. 9.1.1). The crack was forced open and the fracture surface was observed to have been rusted.



Fig. 9.1.1 The segment of the failed fuel pipe in the as-received condition.

Fractographic Examination: The fracture surface was cleaned with acetone ultrasonically and examined under SEM. The sample showed chevron marking over a small lens shaped region near the inner surface (Fig. 9.1.2). The fracture surface contained adherent corrosion debris evidenced by mud cracks (Fig. 9.1.3). EDAX of the corrosion debris revealed the presence of chlorine (Fig. 9.1.4). The rust was cleaned with an oxide remover followed by cleaning with acetone and re-examined. The sample revealed transgranular fracture at the lens shaped region followed by intergranular fracture all through the remaining section thickness (Fig. 9.1.5).

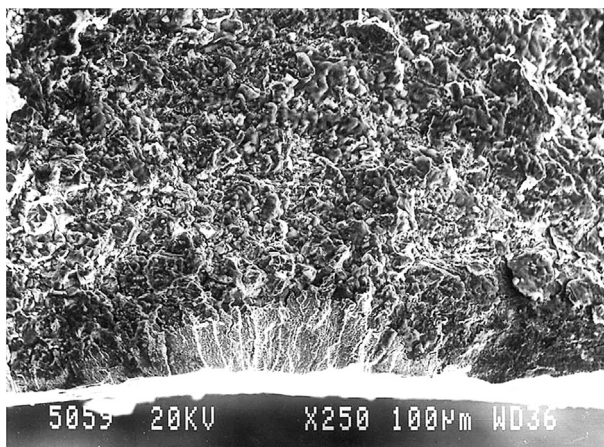


Fig. 9.1.2 SEM fractograph of the fracture near the inner surface showing chevron marking over a small lens shaped region.

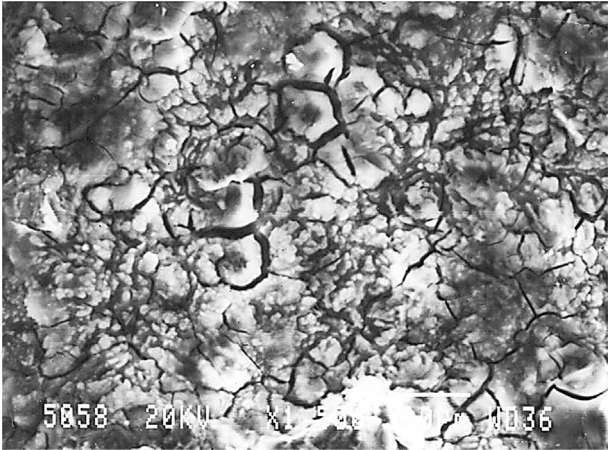


Fig. 9.1.3 SEM fractograph of the fracture surface showing adherent corrosion debris evidenced by mud cracks.

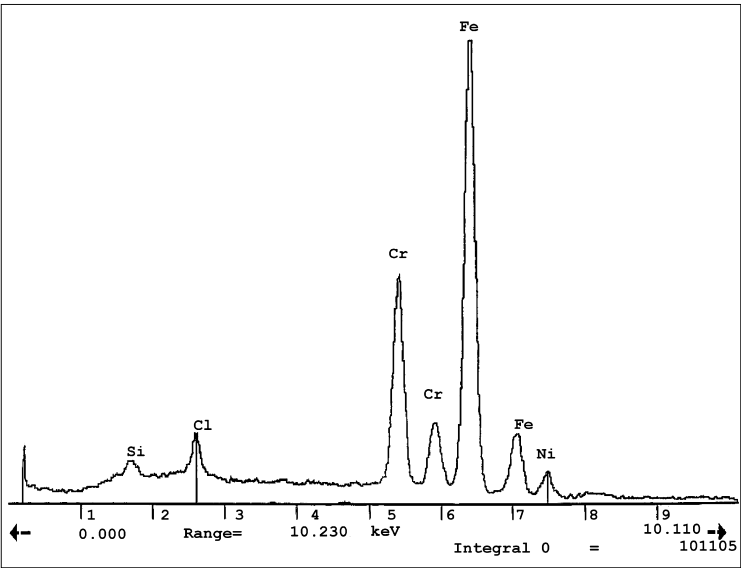


Fig. 9.1.4 EDAX spectrum of the corrosion debris.

Metallographic Examination: Longitudinal section of the sample was prepared for metallography. In the unetched condition the sample showed globular oxide and elongated sulfide inclusions. On etching with aqua regia, the sample revealed equiaxed grains of austenite with annealing twins and

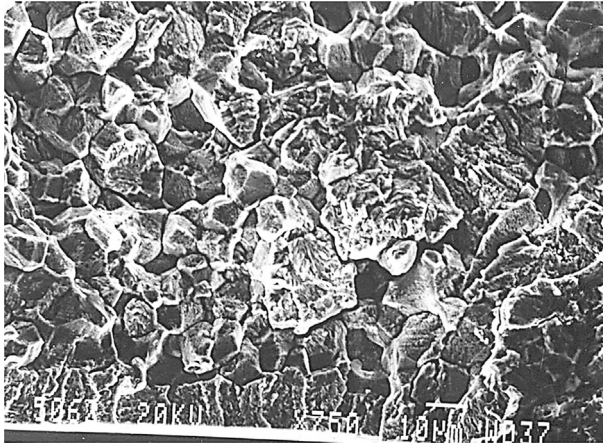


Fig. 9.1.5 SEM fractograph of the larger area of the fracture surface showing intergranular fracture.

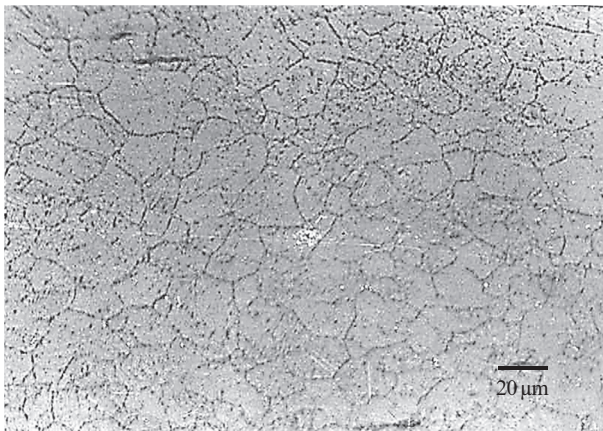


Fig. 9.1.6 Micrograph of longitudinal section of the pipe sample showing equiaxed grains of austenite with annealing twins and chromium carbide precipitates at the grain boundaries. Etchant: Aqua regia.

chromium carbide precipitates at the grain boundaries (Fig. 9.1.6). Mixed trans- and intergranular secondary cracks were observed near the inner surface of the sample (Fig. 9.1.7).

Hardness: Average Vickers hardness of the sample measured at 5 kg load was 195 HV.

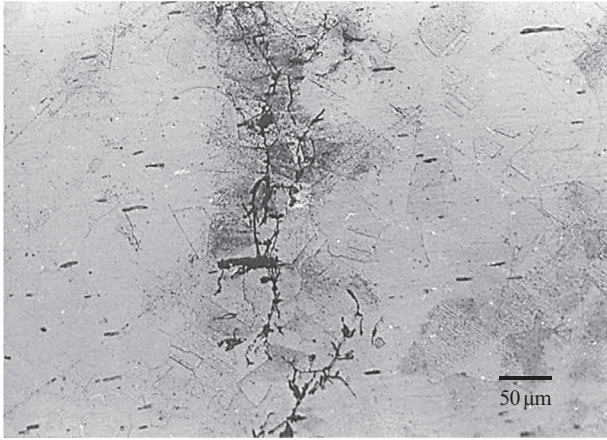


Fig. 9.1.7 Micrograph of longitudinal section of the pipe sample showing a large number of sulfide inclusions and mixed trans- and intergranular secondary cracks near the inner surface of the sample. Etchant: Aqua regia.

9.1.3 Chemical Analysis (Wt.%)

C	Si	Mn	Ni	Cr	S	P	Fe
0.09	0.60	9.0	6.10	16.10	0.14	0.04	Rem.

9.1.4 Discussion

Chemical composition of the steel used for making the nipple did not conform to the specified grade, instead it nearly conformed to grade AISI 202. The sulfur content and the sulfide inclusion population in the steel were too high to satisfy the specification requirements. Metallographic examination indicated that the component was in sensitized condition. In the absence of material processing details, it was difficult to judge at which stage i.e., soon after welding or during cooling after annealing treatment did the material get sensitized. Fractographic examination indicated that the fracture surface was corroded and the debris essentially contained chlorine as evidenced by EDAX. The transgranular fracture at the origin associated with corrosion debris containing chlorine was indicative of crack initiation due to stress corrosion. Mixed trans- and intergranular branching secondary cracks are also evidences of SCC in stainless steels. The intergranular fracture in the remaining fracture surface could be considered to be the result of application of load to a sensitized and corroded material.

9.1.5 Conclusion

The nipple had failed due to poor material quality, a sensitized microstructure and subsequent corrosion from the inner surface, probably by residual chloride-rich water used during the initial pressure testing.



9.2 CASE STUDY: FAILURE OF CONDENSER TUBES OF THERMAL POWER STATION

9.2.1 Background

A number of condenser tubes of a thermal power station developed leaks near the water inlet side of the condenser a day after the unit was recommissioned after overhaul. It was known from the records that the condenser was released for overhaul after it served for 9 months and recommissioned after 6 months of that particular year. The failed condenser tubes were made from cupro-nickel conforming to B.S. 2871-Part 2-1972 grade CuN 102 in annealed condition.

9.2.2 Experimental Results

Visual Examination: A number of small holes of 1–2 mm diameter were observed on the condenser tubes (Fig. 9.2.1). The holes, in general, were found on the bottom half of the tubes. A sample tube was sectioned longitudinally to observe the condition of the inner surface. At a number of locations isolated small pockets of green colored corrosion debris were observed. There were no evidences of erosion ripples.

Metallographic Examination: A longitudinal section containing a hole was prepared for metallography. Silver paint was applied on the inner surface in order to retain the corrosion products. The hole containing the pit at which it originated is shown in Fig. 9.2.2. The pit, which had later developed into a hole, contained dark gray colored corrosion debris (Fig. 9.2.3). On etching, the sample revealed equiaxed grains of alpha and strain markings (Fig. 9.2.4).

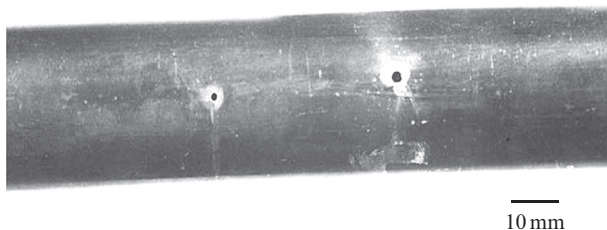


Fig. 9.2.1 A portion of failed condenser tube in the as-received condition.

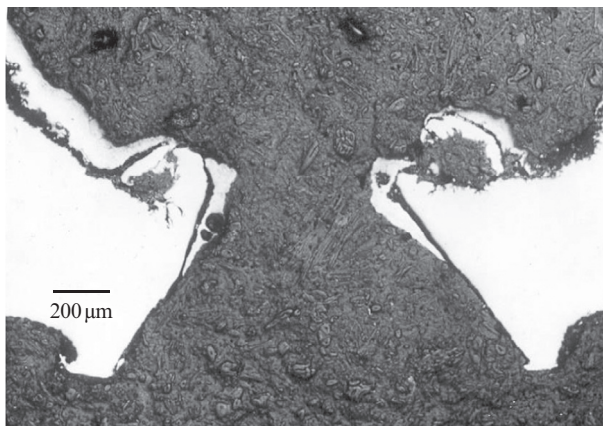


Fig. 9.2.2 Micrograph of the tube sample across the hole. Unetched.

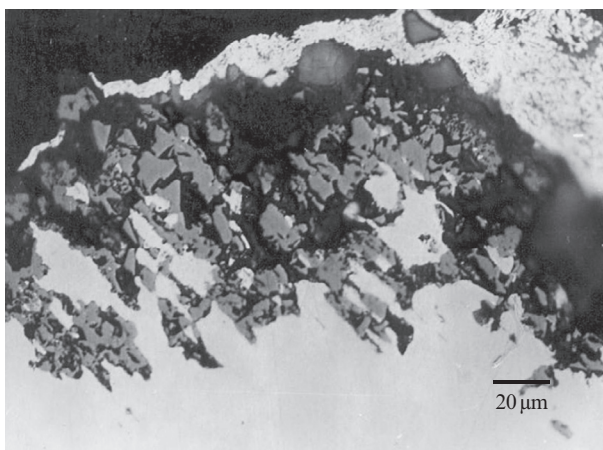


Fig. 9.2.3 Micrograph of the tube sample showing corrosion debris at a pit. Unetched.

Electron Probe Micro Analysis (EPMA): The corrosion debris at the pit shown in Fig. 9.2.3 was analyzed by EPMA. The BSE image along with elemental X-ray mapping for Cu, Ni, Cl, O, Si, Al, and Ag are presented in Fig. 9.2.5. It is evident from the figure that there is depletion of nickel and enrichment of chlorine, oxygen, silicon, and aluminum in the corroded region. Corrosion debris in the Fig. 9.2.5 is below the silver paint.

Hardness: The average hardness of the sample measured at 5 kg load was 75 HV.

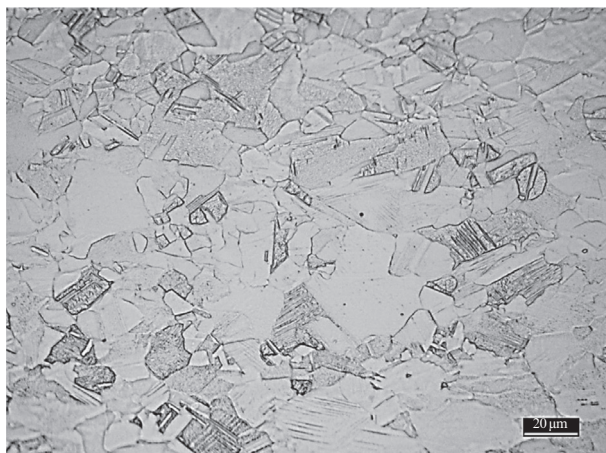


Fig. 9.2.4 Micrograph of longitudinal section of the sample showing equiaxed grains of alpha and strain markings. Etchant: Acidified potassium dichromate.

9.2.3 Chemical Analysis (Wt.%)

Ni	Fe	Mn	Cu
9.80	1.61	0.44	Rem.

9.2.4 Discussion

The chemical composition and hardness conformed to the specified grade of cupro-nickel alloy in annealed condition. Metallographic examination indicated that the tubes had suffered pitting corrosion ultimately resulting in perforation. EPMA revealed that the corrosion was caused by the presence of chloride in the coolant water leading to denickelification. The fact that the tubes were in good condition at the time of shut down and such large number of them developed leaks in a day after recommissioning suggested that the corrosion damage had occurred during the idle period which was about 6 months in this case. Though the users did not mention the conditions of the condenser tubes during the overhaul period, the extent of damage suggested stagnant conditions leading to pitting corrosion.

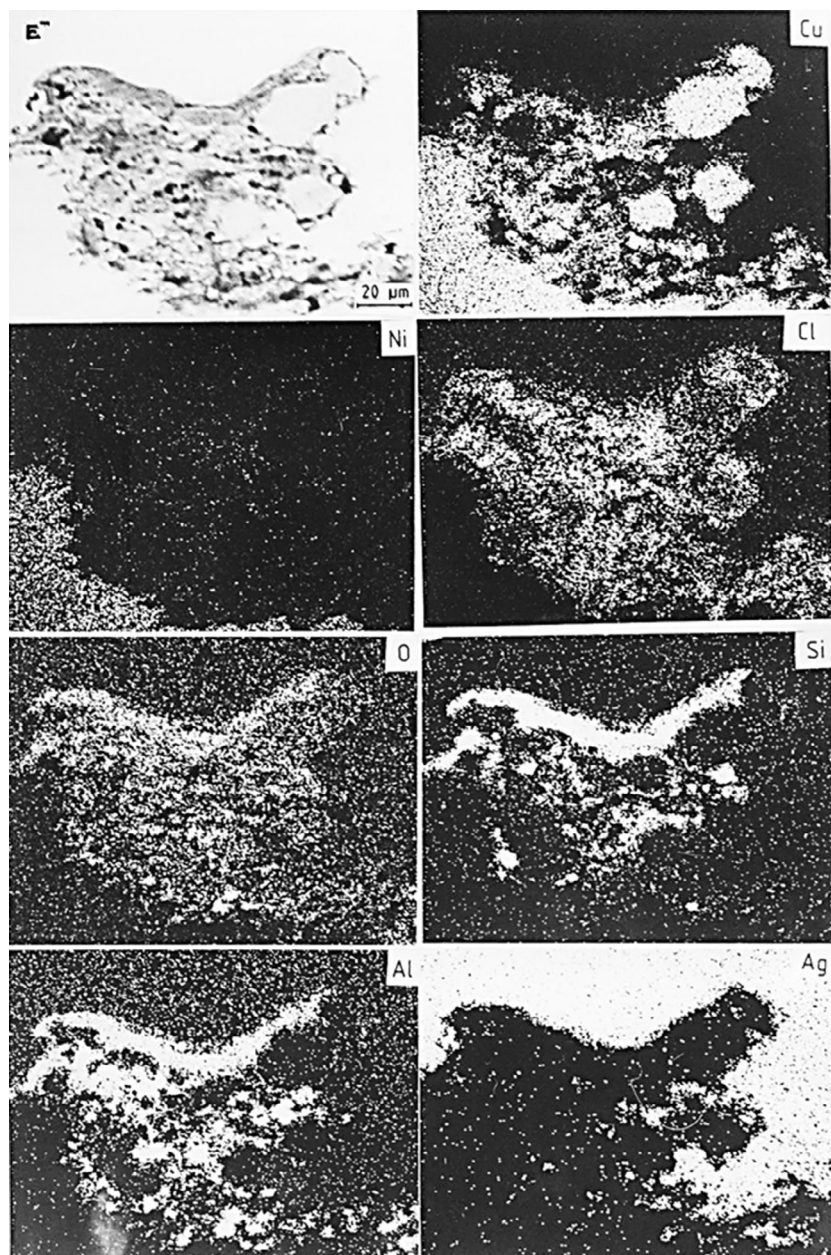


Fig. 9.2.5 The BSE image along with elemental X-ray mapping for Cu, Ni, Cl, O, Si, Al, and Ag.

9.2.5 Conclusion

Pitting corrosion due to chloride-rich coolant water caused the failure of the condenser tube.



9.3 CASE STUDY: FAILURE OF HIGH PRESSURE AIR AND HYDRAULIC PIPES

9.3.1 Background

A few high pressure (H.P) air pipes and hydraulic pipes used in submarine failed prematurely while they were subjected to re-pressure tests in the dry dock after maintenance of the submarine systems. The H.P. air pipes failed at 80 kg/cm^2 whereas the proof pressure was 200 kg/cm^2 . The seamless pipes were made using AISI 321 type austenitic stainless steel and used in an annealed condition.

9.3.2 Experimental Results

Visual Examination: The failed pipe sample in the as-received condition is shown in Fig. 9.3.1. The pipes showed high density of pitting on their outer surfaces, while the inner surfaces of all the pipe samples were free from pitting. The localized pits were aggregated into pockets rather than seen uniformly over the surface.

Metallographic Examination: Longitudinal sections of the pipe samples were prepared for metallographic examination. In the unetched condition, the samples showed a large number of globular oxide and nitride inclusions (Fig. 9.3.2). The outer surface showed a number of surface and subsurface pits (Fig. 9.3.3). The depth of the largest pit observed was $400 \mu\text{m}$. The inner

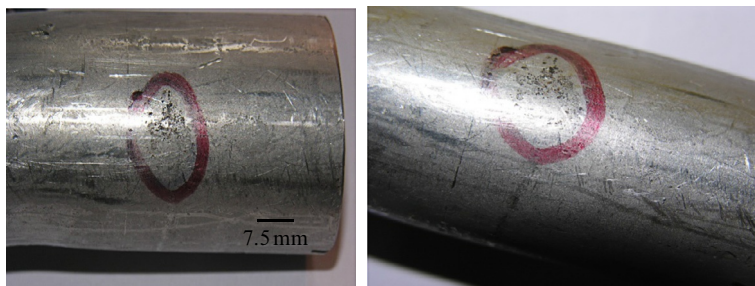


Fig. 9.3.1 Failed pipe samples in the as-received condition

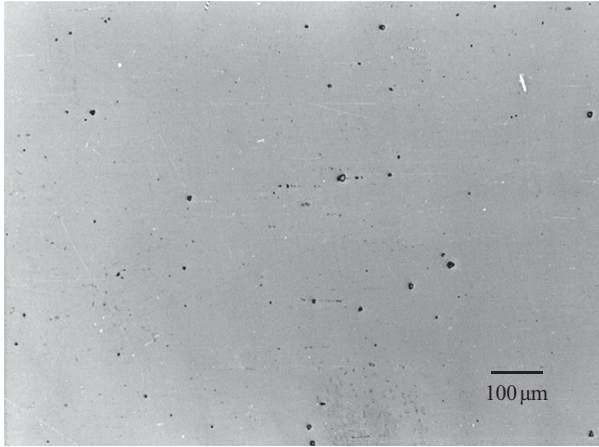


Fig. 9.3.2 Micrograph of longitudinal section of the sample showing a large number of globular oxide and nitride inclusions. Unetched.

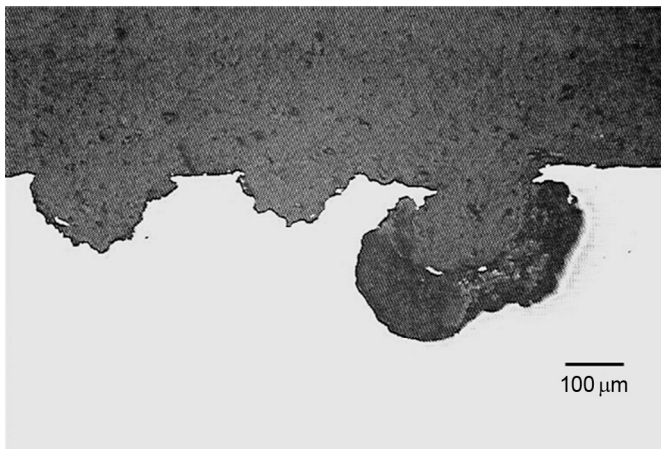


Fig. 9.3.3 Micrograph of longitudinal section of the sample showing a number of pits on the outer surface. Unetched.

surface was free from pits. On etching with aqua regia the sample revealed equiaxed grains of austenite with annealing twins (Fig. 9.3.4).

Energy Dispersive Spectroscopic Analysis (EDAX): EDAX of pitted area containing the corrosion debris showed the presence of sulfur, chlorine, potassium, calcium, and silicon in addition to elements of the parent stainless steel (Fig. 9.3.5).

Hardness: Average Vickers hardness values of the pipe samples measured at 5 kg load were as given here.

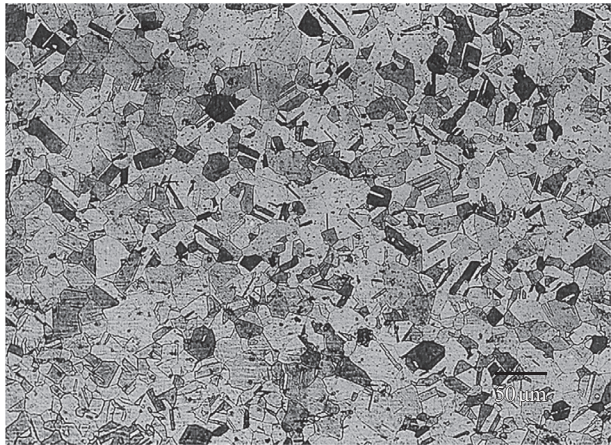


Fig. 9.3.4 Micrograph of longitudinal section of the sample showing equiaxed grains of austenite with annealing twins.

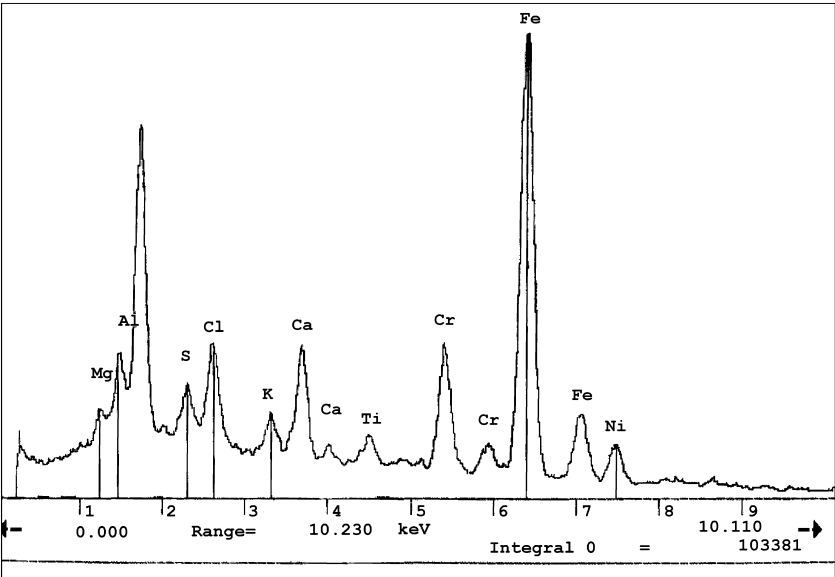


Fig. 9.3.5 EDAX of pitted area showing the presence of sulfur, chlorine, potassium, calcium, and silicon in addition to elements of the parent stainless steel.

Sample	Hardness
H.P. Air pipe	178 HV
H.P. Hydraulic pipe	176 HV

9.3.3 Chemical Analysis (Wt.%)

	C	Si	Mn	Ni	Cr	Ti	Fe
Air pipe	0.075	0.68	1.14	10.80	17.30	0.54	Rem.
Hydraulic pipe	0.081	0.75	1.18	10.80	17.34	0.58	Rem.

9.3.4 Discussion

Chemical analysis of the pipe samples suggested that they were made of titanium-stabilized austenitic stainless steel conforming to type AISI 321. The microstructure and hardness indicated that the pipes were in annealed condition. The pits observed on the outer surface and the presence of elements viz. Ca, K, S, and Cl within the pits reveal that the pipes have suffered pitting corrosion in chlorine and sulfur containing environment. The depth of pitting corrosion was a significant fraction of the wall thickness of the pipes.

9.3.5 Conclusion

The failure of H.P. air and hydraulic pipes during re-pressurization test was caused by extensive pitting corrosion.

9.3.6 Recommendation

- (i) Use AISI 316L or AISI 316T (based on the strength required) that are more pitting corrosion-resistant stainless steels as compared to AISI 321.
- (ii) Do not paint stainless steel pipes as localized damages of painted surface can be sites prone to pitting corrosion.



9.4 CASE STUDY: FAILURE OF DETONATOR OF GRENADE

9.4.1 Background

Defects have been noticed in the detonators used with high explosive (HE) grenades. The detonator has a delay tube machined from aluminum alloy rod conforming to Indian Standards IS:733-1983, designation 24345 in 'W' condition or IS:733-1983, designation 64430 in 'WP' condition. The delay tube is covered with a capsule made of commercially pure aluminum conforming to Indian Standards IS:737-1986, designation 19000 in 'O' condition or IS:739-1977, designation 19500 in 'O' condition. The delay

tube is filled with a delay composition. The chemical composition of the delay filling consists of nitro cellulose, read lead, zirconium powder, and some volatile matter.

The detonator was stated to have been filled with the explosive, assembled, and stored in sealed containers for 2 years before defects were noticed. The observed defects were formation of white encrustation on the filled delay tubes, spillage of the filling and bulging of the tubes in certain cases. The delay tube containing an explosive filling comprised of lead oxide, antimony sulfide, barium chromate, and potassium perchlorate. As per the users, the composition of the explosive filling was compatible with the aluminum alloy used for making the delay tube.

A few samples of both the defective assembled delay tube with capsule and also unassembled ones without filling for comparative study were received for investigation.

9.4.2 Experimental Results

Visual Examination: The delay tube along with the capsule, of both assembled (defective) as well as unassembled, is shown in Fig. 9.4.1A and B, respectively. The defective tube samples showed thick adherent grayish white colored corrosion debris on the outer surface. On cutting the sample longitudinally into two halves the inner surface also showed thick adherent gray colored corrosion debris. The machined outer and inner surfaces of the unassembled tube sample showed metallic luster. The capsule of the failed tube was slightly bulged and showed loosely held grayish green deposit on its inner surface.

Metallographic Examination: Longitudinal and transverse sections of the defective delay tube, the unassembled delay tube, and the capsule samples were prepared for metallographic examination. In the unetched condition, both the longitudinal and the transverse sections of the defective tube samples showed pitting corrosion and intergranular cracks on the outer as well as the inner surfaces (Figs. 9.4.2–9.4.5). Similar defects were not noticed on the inner and the outer surfaces of the unassembled tube sample. The capsule sample did not reveal any defect in the unetched condition. On etching with Keller's reagent, longitudinal section of the defective tube sample revealed coarse elongated grains of aluminum solid solution with large number of undissolved particles predominantly along the grain boundaries (Figs. 9.4.6 and 9.4.7). The unassembled tube sample showed finer grain structure of aluminum solid solution with finer and relatively fewer amounts

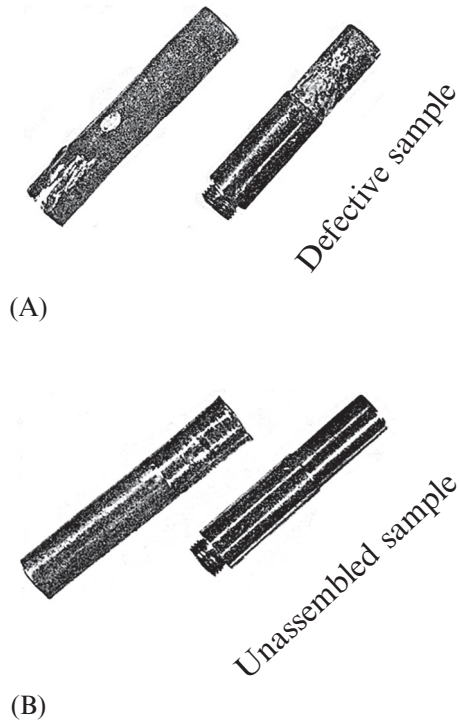


Fig. 9.4.1 Delay tube and capsule samples in the as-received condition. (A) Defective sample showing corroded surface of the delay tube and tool marks on the capsule, (B) Unassembled sample showing machined bright surface.

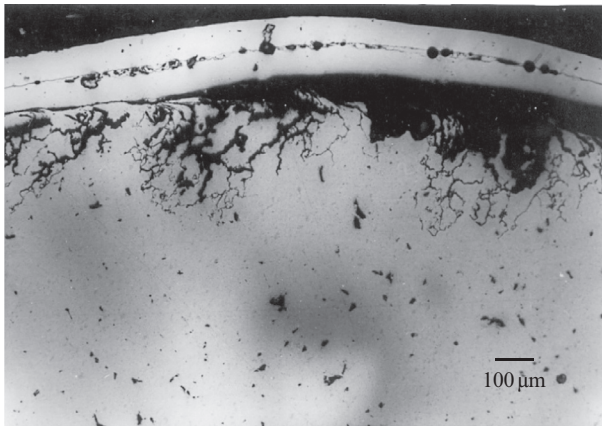


Fig. 9.4.2 Micrograph of transverse section of the defective sample showing pitting and intergranular corrosion on the outer surface. Copper strip was wrapped to protect the outer surface for sample preparation. Unetched.

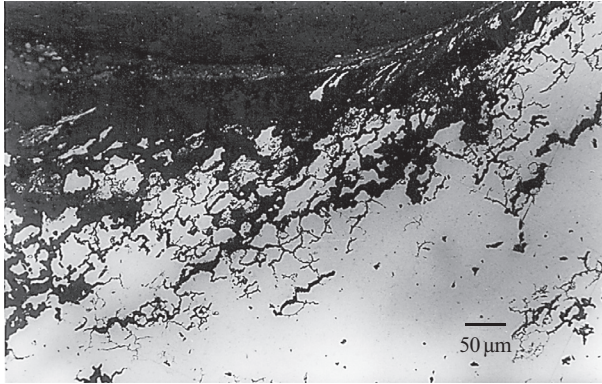


Fig. 9.4.3 Micrograph of transverse section of the defective sample showing intergranular corrosion of the inner surface. Unetched.

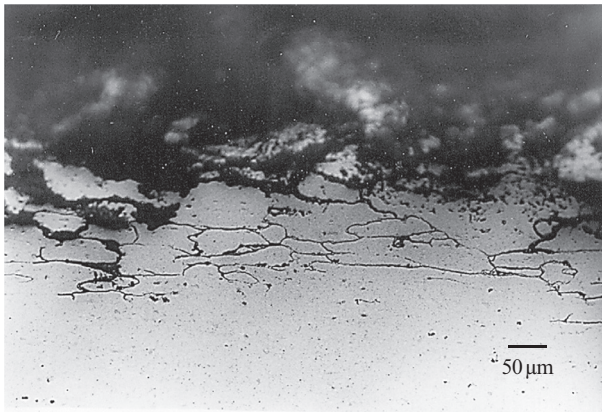


Fig. 9.4.4 Micrograph of longitudinal section of the defective sample showing intergranular corrosion of the outer surface. Unetched.

of undissolved precipitates (Fig. 9.4.8). A section of the capsule sample showed insoluble particles in a matrix of aluminum (Fig. 9.4.9).

Electron Probe Micro Analysis (EPMA): Transverse section of the defective delay tube sample was prepared for EPMA. On examining, the defective tube sample revealed the presence of aluminum oxide layer on the outer and inner surfaces and also the grain boundaries near corroded region of both the surfaces. Backscattered electron image and X-ray elemental mappings for aluminum and oxygen of the defective sample are shown in Fig. 9.4.10.

EDS Analysis: The corrosion debris present on the inner surface of the defective delay tube sample was subjected to energy dispersive spectroscopic

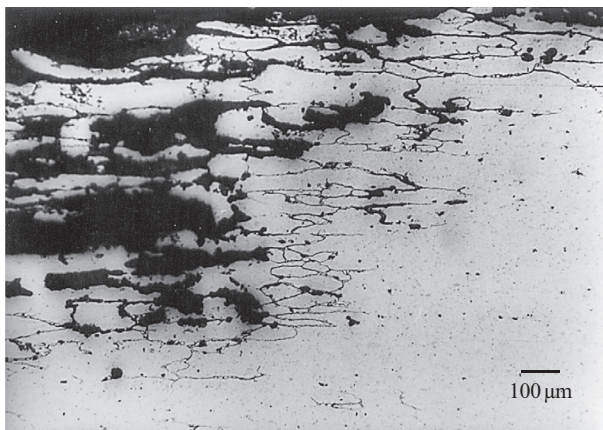


Fig. 9.4.5 Micrograph of longitudinal section of the defective sample showing intergranular corrosion of the inner surface. Unetched.



Fig. 9.4.6 Micrograph of longitudinal section of the defective tube sample showing large number of undissolved particles and grain boundary precipitation of CuAl_2 near the corroded surface. Etchant: Keller's reagent.

(EDS) analysis. The overall and the point analysis of the corrosion debris on the defective delay tube sample showed the presence of oxygen in addition to the constituent elements of the parent material and the explosive filling. EDS spectrum, SEM micrograph of the analyzed region, and the results of point analysis of the defective delay tube sample are shown in [Fig. 9.4.11](#).

Hardness: Average Vickers hardness values of the samples measured at 5kg load were as given here.

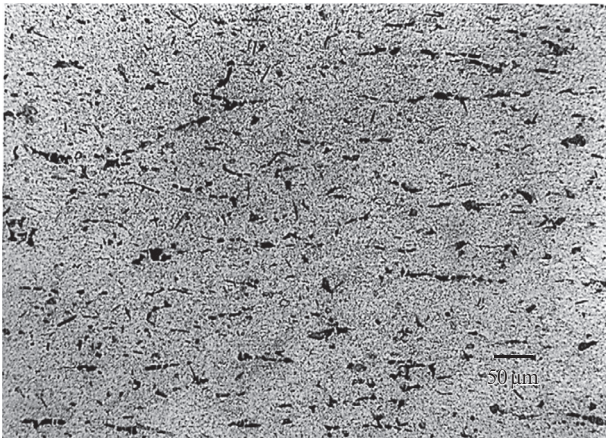


Fig. 9.4.7 Micrograph of the longitudinal section of the defective tube sample showing large number of undissolved particles and grain boundary precipitation of CuAl₂ away from the corroded surface. Etchant: Keller’s reagent.

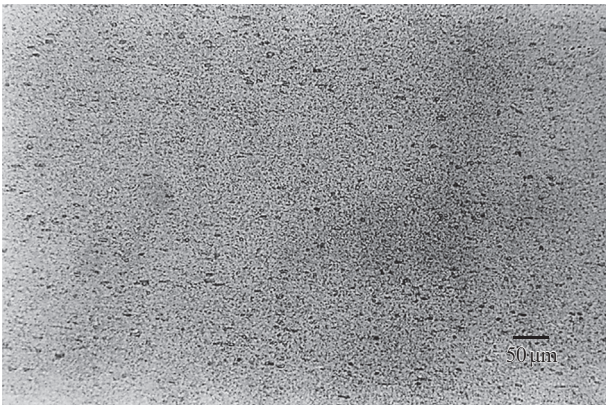


Fig. 9.4.8 Micrograph of longitudinal section of the unassembled tube sample showing relatively finer undissolved particles in a matrix of aluminum solid solution. Etchant: Keller’s reagent.

Sample	Hardness (HV)
(i) Defective delay tube	70
(ii) Unassembled delay tube	102
(iii) Capsule of defective tube	35

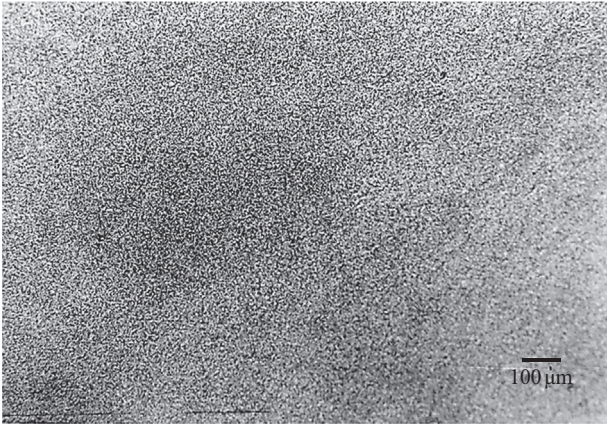


Fig. 9.4.9 Micrograph of longitudinal section of the capsule sample showing fine undissolved particles in a matrix of aluminum. Etchant: Keller’s reagent.

9.4.3 Chemical Analysis (Wt.%)

Chemical composition of the defective delay tube, the capsule, and the unassembled delay tube samples were as given here.

	Cu	Mn	Si	Mg	Fe	Cr	Al
Defective tube	3.95	0.61	0.62	0.38	0.20	0.23	Bal
Capsule	0.015	0.005	0.23	0.23	0.20	—	Bal
Unassembled tube	0.07	0.52	0.92	1.06	0.24	0.05	Bal

9.4.4 Discussion

Chemical analysis of the defective delay tube sample suggested that the component was made using the specified grade (24345, IS:733–1983) of aluminum alloy. Microstructural examination revealed a large number of coarse undissolved particles and grain boundary precipitation of CuAl_2 particles. The observation indicated that the material of delay tube was not properly heat treated. Inadequate solution treatment parameters employed have caused insufficient dissolution of the particles in the aluminum alloy. This has resulted in a microstructure, which has large population of undissolved particles rendering it both inhomogeneous and a solid solution of very low strength. Lower strength of the material was also confirmed by its very low hardness (70 HV) as compared to the minimum hardness (110HV) for grade 24345 alloy in the specified ‘W’ (solution treated) condition. Moreover,

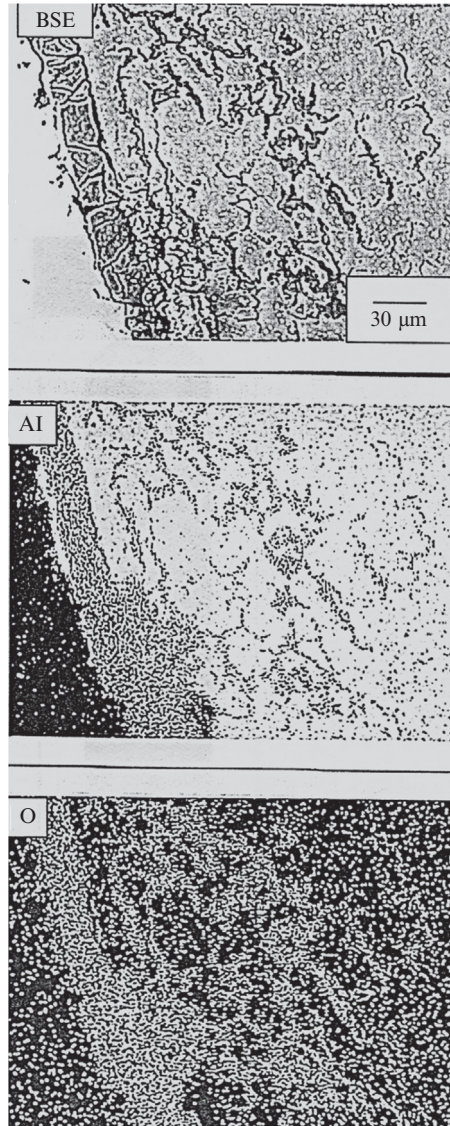
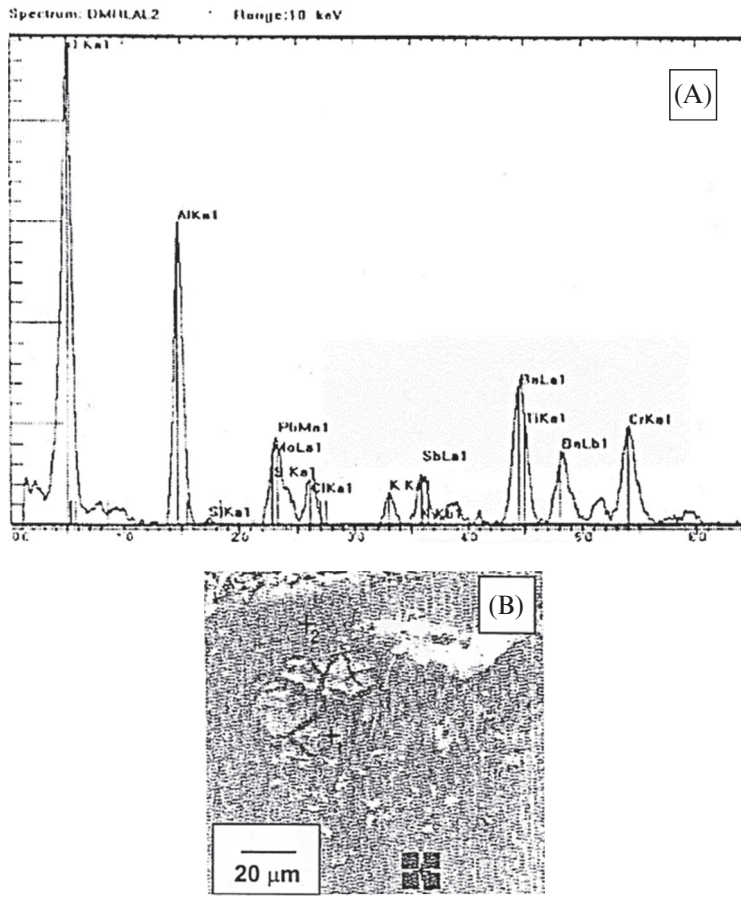


Fig. 9.4.10 BSE image of the transverse section of the defective tube sample showing corroded outer surface and grain boundaries along with X-ray elemental mapping for Al and O.

delay caused during quenching or low rate of cooling during quenching has allowed precipitation of CuAl_2 particles along the grain boundaries resulting in impoverishment of copper from the vicinity of the grain boundaries rendering the alloy susceptible to intergranular corrosion.



	O(Ka)	Al(Ka)	Cl(Ka)	K(Ka)	Ti(Ka)	Cr(Ka)	Mo(La)	Sb(La)	Ba(La)	Pb(Ma)
1	52.09	15.26	1.66	1.20	0.89	4.14	5.38	4.05	10.66	4.68
2	45.09	16.49	1.56	1.20	0.62	4.95	4.76	5.68	14.26	5.38
3	27.91	4.87	0.23	0.18	1.25	10.49	6.67	11.19	26.25	10.96

Fig. 9.4.11 EDS analysis of the corrosion debris on the inner surface of the defective delay tube.

Microstructural examination has revealed extensive intergranular corrosion of the defective delay tube samples near the outer as well as the inner surfaces. The corrosion debris mainly comprised of aluminum oxide, as confirmed by EPMA and aluminum and oxygen along with the constituents of the explosive filling, as confirmed by the EDS analysis. The delay tube containing an explosive filling in a capsule seemed to have a conducive

atmosphere for corrosion of the tube material which already had a highly susceptible microstructure.

Even though the chemical composition of the delay tube conformed to the specification, it would remain compatible to the explosive filling only in the properly heat-treated condition. Improper heat treatment renders the microstructure highly susceptible to corrosion and not so compatible to the explosive filling any more, as observed in the present case.

Chemical analysis of the unassembled delay tube sample suggested that it was made using another specified grade (64430, IS:733-1983) of aluminum alloy. As compared to the defective delay tube samples the unassembled ones showed relatively homogenous microstructure that contained finer and slightly lower population of undissolved particles. Although the hardness of the unassembled tube sample was higher than the defective tube, it was lower than that is specified (105 HV minimum) for the alloy in 'WP' (solution treated and precipitation hardened) condition. The result suggested that the unassembled tube samples were also not solution treated and precipitation hardened adequately. However, the unassembled sample was not of the same composition and therefore it is expected to show slight difference in microstructure and hardness after processing and heat treatment.

Chemical analysis, microstructure, and hardness of the capsule sample suggested that it was made of the specified grade (19000, IS:737-1986) of commercially pure aluminum in annealed condition.

9.4.5 Conclusion

The delay tubes had suffered intergranular corrosion of their surfaces due to improper heat treatment of the material used, rendering them susceptible to corrosion.



Hydrogen Embrittlement and Stress Corrosion Cracking

Hydrogen embrittlement is a failure process wherein absorption of excessive amounts of hydrogen in a metal together with the presence of residual or applied tensile stress causes the failure. Most frequently, it is found that high strength steels and other high strength alloys suffer from hydrogen embrittlement. Hydrogen embrittlement causes lack of tensile ductility in metals and alloys and leads to cracking called the hydrogen cracking. The sources of hydrogen in the environment could be water, moist air, acids, hydrogen sulfide, hydrocarbons, etc.

One important characteristic of hydrogen cracking is a specific delay time for appearance of cracks after stress is applied. Hydrogen embrittlement can be prevented by the following methods:

- (a) Using clean steels. Killed steels increase resistance of hydrogen cracking in relation to rimmed steels.
- (b) Using liners or coatings impervious to hydrogen
- (c) Using inhibitors
- (d) Baking after any kind of plating
- (e) Substituting alloys with less susceptible ones

Simultaneous presence of tensile stress and a specific corrosive medium cause stress corrosion cracking (SCC) in metals and alloys. Pure metals are resistant to SCC whereas alloys are susceptible to it. The stress may be applied, residual or thermal in nature and can be lower than the yield stress. For a given alloy–environment combination, there is a minimum value of stress called the threshold stress below which SCC does not occur. The environment causing the SCC is generally specific for a specific alloy system. For example, brasses crack in ammonia-rich environment and not in chlorides whereas stainless steels behave the opposite. The cracking pattern of SCC may be intergranular, transgranular, or mixed.

The following preventive methods may prove to be beneficial in combating SCC.

- (a) Lowering the stress below the threshold value. This can be achieved by removing the residual stress in a component.
- (b) Cladding an alloy with a pure metal, for example, cladding aluminum alloy with pure aluminum.

- (c) Changing the alloy if environment and stress cannot be changed.
- (d) Applying cathodic protection.
- (e) Adding inhibitors to the system if feasible.



10.1 CASE STUDY: FAILURE OF FIN MOUNTING SCREWS OF AN AERO WEAPON

10.1.1 Background

An aero weapon uses M8 screws and hexagonal lock nuts for fastening of the fins. Each fin requires four sets of screws and nuts. The torque used to fasten the screw was 20Nm. In one such incident while torque was being applied two screws got cracked. The screws were made of a low alloy Ni-Cr-Mo steel of grade SAE 4340, hardened at 860°C and tempered at 550°C. The screws were cadmium plated and baked.

10.1.2 Experimental Results

Visual Examination: The failed screw in the as-received condition is shown in Fig. 10.1.1. The screw was found to have broken along the edge of the central slot. The fracture surface was coarse and crystalline.



Fig. 10.1.1 The failed screw in the as-received condition.

Fractographic Examination: The fracture surface was cleaned ultrasonically in acetone and examined under scanning electron microscope (SEM). The fracture surface revealed essentially intergranular fracture (Fig. 10.1.2).

Metallographic Examination: Longitudinal section of the screw showed large population of globular oxide and slag-like coarse inclusions in the unetched condition (Fig. 10.1.3). On etching with 2% nital, the sample

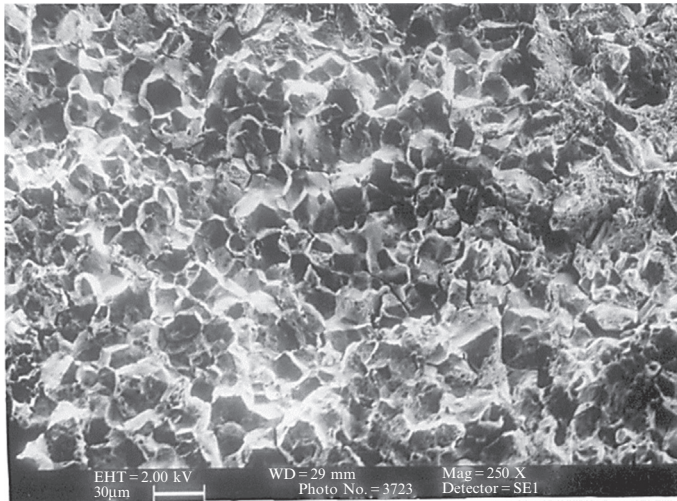


Fig. 10.1.2 SEM fractograph of the fracture surface showing intergranular fracture.

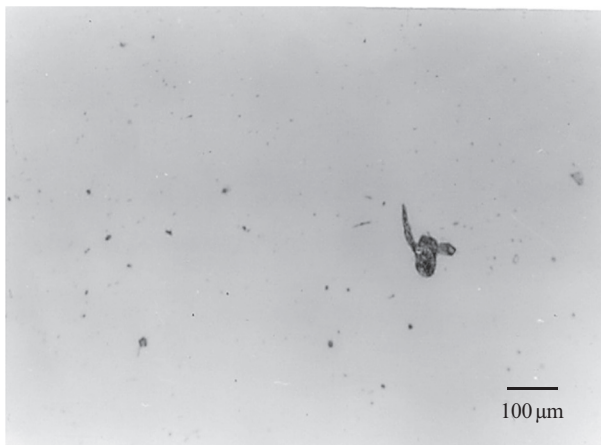


Fig. 10.1.3 Micrograph of longitudinal section of the screw showing large population of globular oxide and slag-like coarse inclusions. Unetched.



Fig. 10.1.4 Micrograph of longitudinal section of the screw showing a tempered martensite structure. Etchant: 2% nital.

revealed a tempered martensite structure (Fig. 10.1.4). Spot test confirmed the presence of cadmium plating on the surface of the screw.

Hardness: Average Vickers hardness of the failed screw at 20 kg load was 428 HV.

10.1.3 Chemical Analysis (Wt.%)

C	Si	Mn	Ni	Cr	Mo	S	P	As, Sb, Sn	Fe
0.40	0.42	0.71	1.60	1.20	0.23	0.025	0.02	<100ppm	Rem.

10.1.4 Discussion

The chemical composition, microstructure, and hardness suggested that the screw was made of Ni-Cr-Mo steel nearly conforming to the specified grade and used in a hardened and tempered condition. As the screw had failed in an intergranular mode, the reasons for the failure could be,

- (a) Quench cracking
- (b) High temperature service
- (c) Temper embrittlement or
- (d) Hydrogen embrittlement

The screw was used at ambient temperature so the high temperature service causing the failure can be eliminated. The fracture surface did not

show any evidence of seeping in of plating solution which would have resulted in the event of a preexisting quench crack before plating. Thus quench cracking cannot be the reason for failure. The low content of embrittling elements viz., As, Sb, and Sn in the composition and presence of Mo to scavenge phosphorous segregation to prior austenite grain boundaries, eliminate the possibility of temper embrittlement of the steel. Although the steel was baked after cadmium plating it could be doubted that either the baking parameters i.e., temperature and time were not adequate for complete flushing of hydrogen from the component or there could be delay between cadmium plating and baking resulting in hydrogen embrittlement of the screw.

10.1.5 Conclusion

The most probable cause of failure was hydrogen embrittlement of the screw.



10.2 CASE STUDY: FAILURE OF DIFFERENTIAL VALVE COVER OF FUEL METERING UNIT

10.2.1 Background

The differential valve cover with a collar pin of a fuel metering unit was found broken. The differential valve cover/bracket is one of the three brackets with which the fuel metering unit is mounted on the engine. The fuel metering unit regulates fuel feed into the engine combustion chamber at steady and transient power settings. Failure of the differential valve cover can cause drooping of fuel metering unit and thereby affecting low-pressure rotor RPM regulation. The life of the component before failure was 301:50 hours against a service life of 1600 h.

10.2.2 Experimental Results

Visual Examination: The broken differential valve cover/bracket in the as-received condition is shown in [Fig. 10.2.1](#). The component is found to have broken from an area, near the pin. The area near the failure did not show any plastic deformation. The fracture surface was cleaned and observed under stereo microscope. The surface showed a coarse and crystalline fracture. Photograph of the close-up view of the mating fracture surfaces is shown in [Fig. 10.2.2](#).

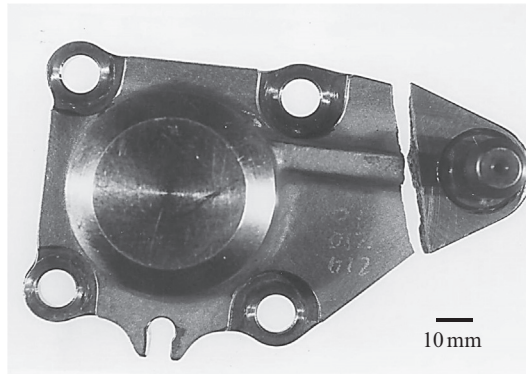


Fig. 10.2.1 Broken differential valve cover in the as-received condition.

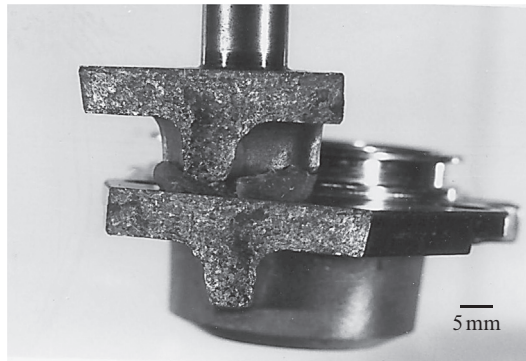


Fig. 10.2.2 Fracture surfaces of the component showing coarse and crystalline fracture.

Fractographic Examination: The mating fracture surfaces were cleaned ultrasonically in acetone and examined under scanning electron microscope (SEM). The fracture surface showed a cleavage fracture with transgranular secondary cracks all over the region without any specific fracture origin (Fig. 10.2.3).

Metallographic Examination: A section of the broken unit was prepared for metallographic examination. In the unetched condition, the sample showed a number of microporosities (Fig. 10.2.4). On etching with Kroll's reagent the sample revealed a transformed beta structure containing plates of acicular alpha (white) and thin plates of beta (dark) (Fig. 10.2.5). Alpha platelets were also seen at prior beta grain boundaries which look like a continuous film of alpha at certain grain boundaries (Fig. 10.2.5).

Hardness: Average Vickers hardness at 30 kg load was 358 HV.

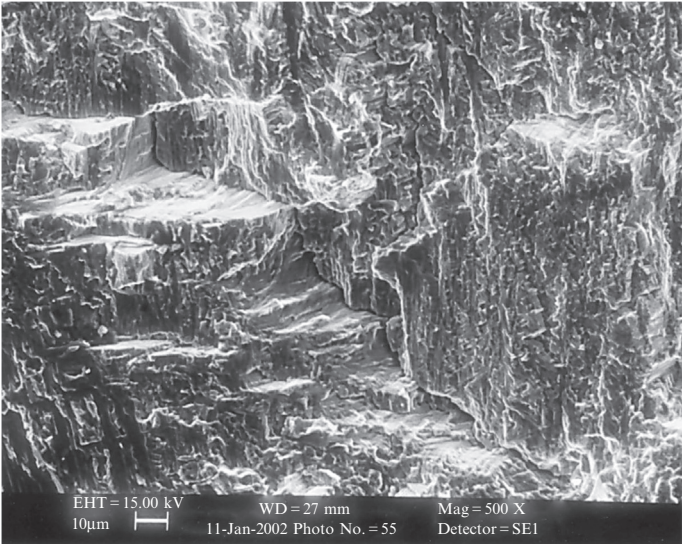


Fig. 10.2.3 SEM fractograph of the fracture surface showing cleavage steps and transgranular secondary cracks.

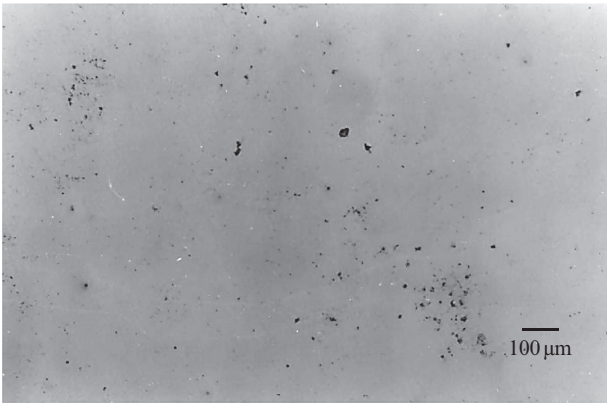


Fig. 10.2.4 Micrograph showing microporosities in the sample. Unetched.

10.2.3 Chemical Analysis (Wt.%)

Al	Mo	Fe	H	O	N	Ti
6.19	1.80	0.51	0.026	0.185	0.0075	Base

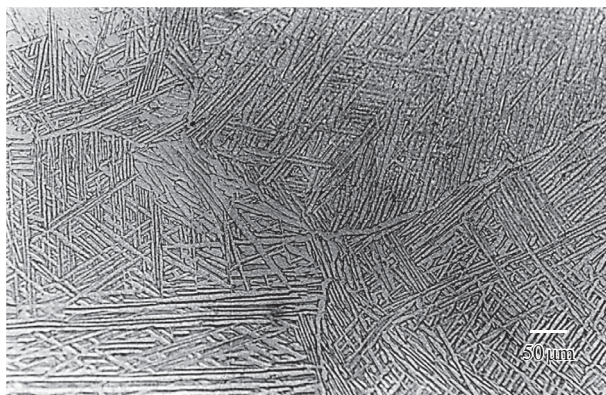


Fig. 10.2.5 Micrograph of the sample showing a transformed beta structure containing acicular plates of alpha and thin plates of beta. Grain boundaries are showing plates and film of alpha. Etchant: Kroll's reagent.

10.2.4 Discussion

The chemical composition, microstructure, and hardness suggested that the differential valve cover/bracket was made of an alpha-beta (α - β) titanium alloy in solution treated and aged condition. The chemical composition revealed that the gaseous content, both hydrogen and oxygen in the alloy, was quite high.

Titanium alloys containing either an α phase or an $\alpha + \beta$ phase and hydrogen in amounts above a certain 'critical' level, which is specified as 0.015% are subject to delayed cracking, irrespective of whether they also contain a hydride phase or not [1]. With passage of time, the high content of hydrogen decreases the resistance of the alloy to delayed cracking. The stresses produced during loading of the component considerably accelerate the process of cracking. In addition to this the tendency of titanium alloys to the delayed cracking is greatly influenced by the ductility of the metal. Alloys with a high content of oxygen and nitrogen, as in the present case, have lower ductility which promotes cracking. Residual stress present in the material can also enhance the stress component reducing the resistance of the alloy to delayed cracking. In the differential valve cover the insertion of a pin at one end causes some amount of residual stress in the component adding to the applied stress.

10.2.5 Conclusion

The differential valve cover had failed due to delayed cracking which was caused by a higher content of hydrogen in the alloy which was more than

that is permissible (0.015% by weight) in such alloys for engineering applications.

10.2.6 Reference

- [1] A.S. Mikhailov, B.S. Krylov, 'The Influence of Hydrogen on the Tendency of Titanium Alloys to Delayed Cracking', Titanium and its Alloys, Ed. 1.1. Kornilov, Publication No. 10 Translated from Russian, Israel Program for Scientific Translation, Jerusalem, 1966, pp. 154–157.



10.3 CASE STUDY: FAILURE OF LEVER ASSEMBLY OF MAIN UNDER CARRIAGE OF AN AIRCRAFT

10.3.1 Background

During post flight inspection it was noticed that the lever assembly of forward door operating system in the main under carriage of an aircraft had failed. It is also reported that in the past few years, similar failures were noticed during retraction of the under carriage in flying or ground run. The failed component was an aluminum alloy casting conforming to grade BS 4L53.

10.3.2 Experimental Results

Visual Examination: The failed lever assembly in the as-received condition is shown in Fig. 10.3.1. The component was an integral piece of three arms with steel inserts at all the three ends which were secured in position with rivets. The component was found to have been painted. The component



Fig. 10.3.1 The failed lever assembly in the as-received condition.

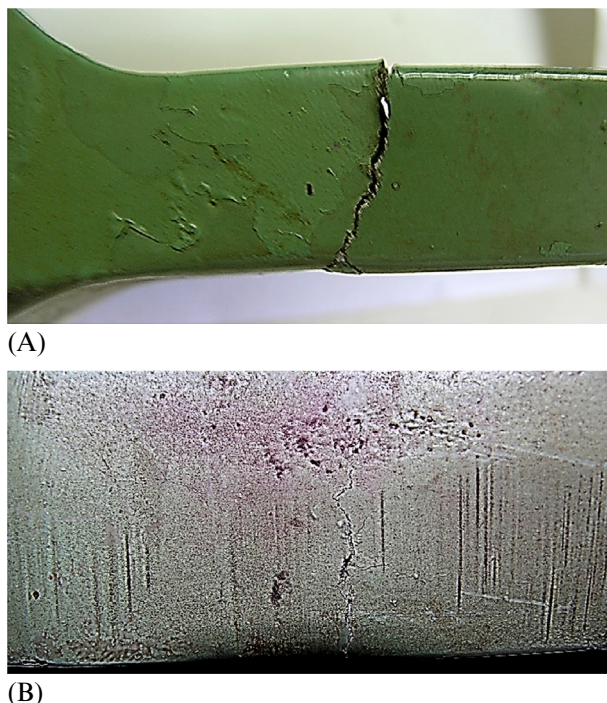


Fig. 10.3.2 (A) Close-up view of the crack as mentioned in [Fig. 10.3.1](#). (B) Pits and deep grooves on the inner surface of the component.

showed a breakage on one of the arms and a crack on another arm. A close-up view of the crack is shown in [Fig. 10.3.2](#). The inner surface of the component was found to be rough and soiled and showed pits and deep grooves. The fracture surface was coarse and crystalline.

Fractographic Examination: The fracture surfaces from both broken and cracked regions were cleaned ultrasonically in acetone and examined under scanning electron microscope (SEM). Near the surface the fracture surface showed pockets of intergranular fracture and dendritic features ([Fig. 10.3.3](#)), while in the central region it showed quasi-cleavage fracture with pockets of dimpled rupture ([Fig. 10.3.4](#)).

Metallographic Examination: Longitudinal and transverse sections of samples taken from near the broken and cracked regions were prepared for metallographic examination. Section near the surface showed pitting and intergranular corrosion ([Fig. 10.3.5](#)). On etching with Keller's reagent the sample revealed intergranular branching cracks perpendicular to the fracture surface ([Fig. 10.3.6](#)). The general microstructure of the sample consisted

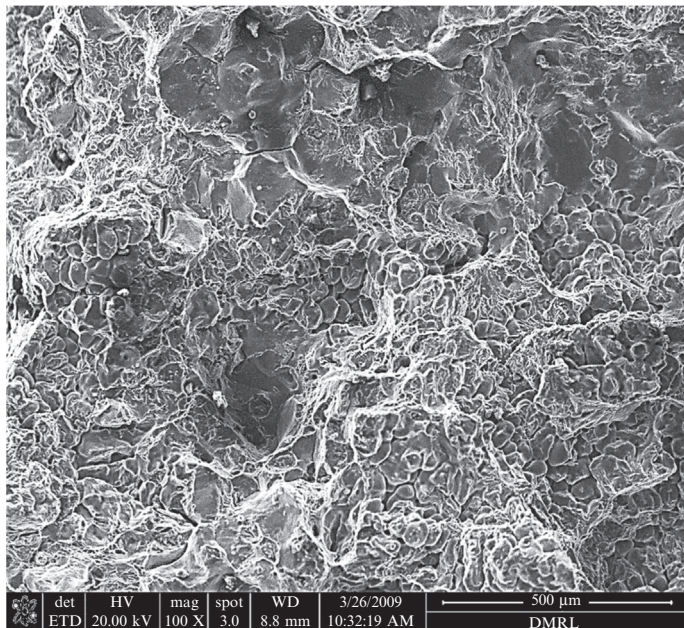


Fig. 10.3.3 SEM fractograph of the fracture surface near the surface showing intergranular fracture features in pockets and also dendritic features.

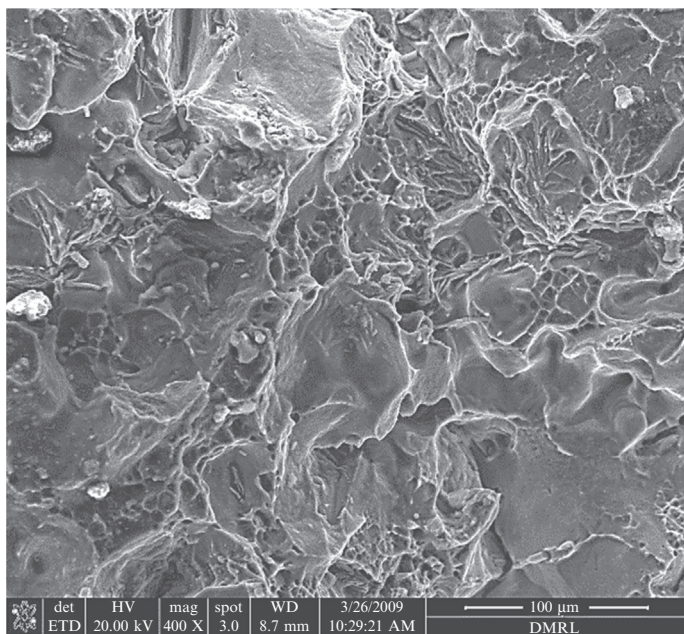


Fig. 10.3.4 SEM fractograph of the fracture surface in the center showing quasi-cleavage fracture and pockets of dimpled rupture.

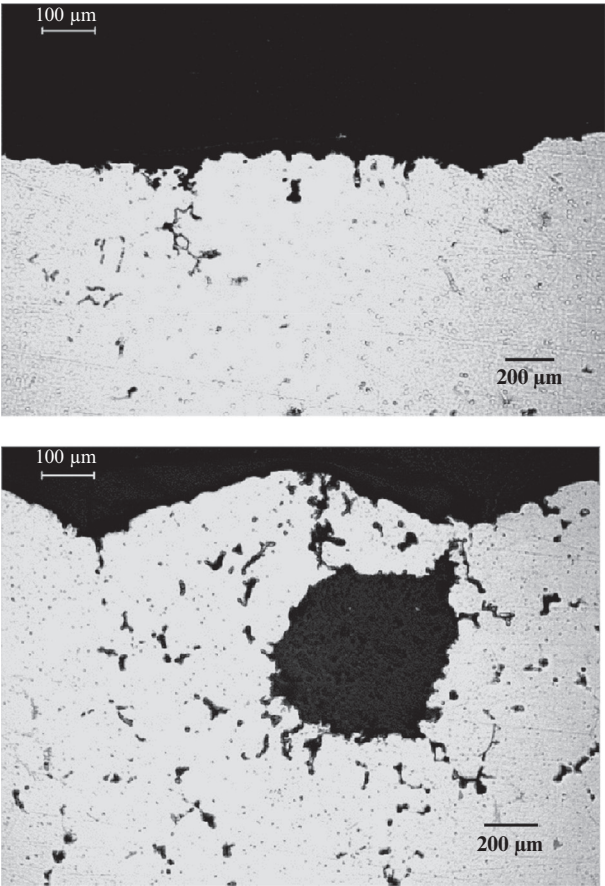


Fig. 10.3.5 Micrograph of a section of the component showing pitting and intergranular corrosion near the surface. Unetched.

of coarse grains of aluminum solid solution and interdendritic network of Mg_2Al_3 phase (Fig. 10.3.7).

Hardness: Average Vickers hardness measured at 10kg load was 110HV.

10.3.3 Chemical Analysis (Wt.%)

Si	Mg	Fe	Al
0.10	9.39	0.12	Base

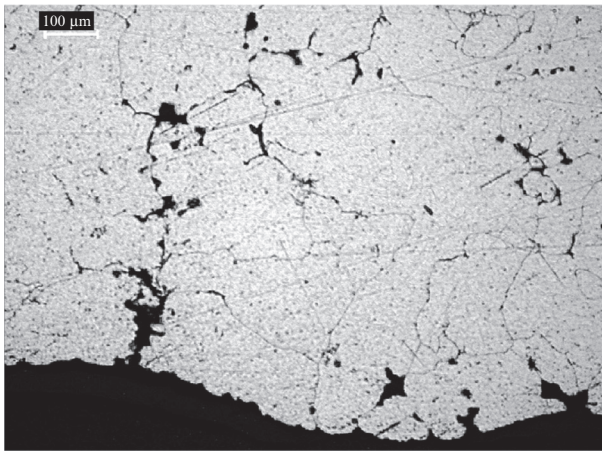
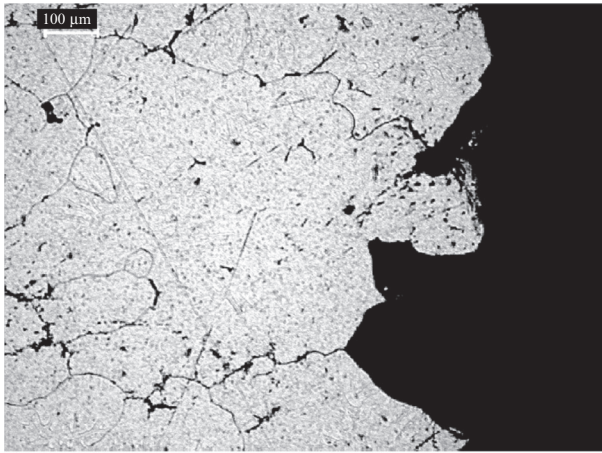


Fig. 10.3.6 Micrograph of the sample showing secondary intergranular branching cracks perpendicular to the primary fracture. Etchant: Keller's reagent.

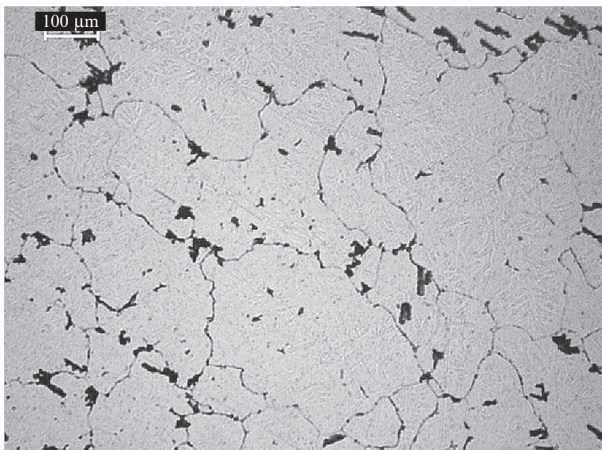


Fig. 10.3.7 Micrograph of the sample showing coarse grains of aluminum solid solution and interdendritic network of Mg_2Al_3 phase (dark). Etchant: Keller's reagent.

10.3.4 Discussion

The chemical composition of the lever assembly suggested that it was made of an aluminum alloy nearly equivalent to specification BS 4L53. Microstructure indicated that it was a casting and hardness suggested that it was used in a solution treated and aged condition.

The microstructure of the component near the failed region revealed surface pitting and intergranular corrosion. The corrosion might have initiated at sites containing deep grooves on the surface. Aluminum alloys with $>3\%$ Mg are susceptible to stress corrosion cracking (SCC) in hot and humid atmosphere. The Mg_2Al_3 precipitates along the grain boundaries get dissolved preferentially in humid or mild corrosive environment causing intergranular corrosion. Tensile stress either residual or operational experienced by the component can cause stress corrosion cracking. It is well known that Al-10Mg alloy casting must be slowly quenched from the solution treatment temperature, otherwise residual stresses may lead to SCC. In addition, the alloy tends to be unstable, particularly in tropical conditions, leading to precipitation of Mg_2Al_3 in grain boundaries, which lowers ductility and may cause SCC after a period of time [1].

The evidences found during analysis point to the fact that lever assembly had developed a crack due to SCC which propagated in intergranular mode to certain depth along the cross section. The remaining section fractured under overload condition due to the operational load.

10.3.5 Conclusion

The lever assembly had failed due to stress corrosion cracking.

10.3.6 Reference

- [1] I.J. Polmear, Cast Aluminium alloys, Metallurgy of Light Metals, 2nd ed., Metallurgy and Material Science series, Edward Arnold, London, 1989, pp 145-168.



10.4 CASE STUDY: FAILURE OF CARTRIDGE SEAT EJECTION SECONDARY OF AN AIRCRAFT

10.4.1 Background

Cartridge seat ejection (CSE) system is used to save the life of the pilot in emergency. These cartridges are installed on the telescope tube, which is located behind the seat. The cartridge secondary consists of brass case, cap

(with highly sensitive composition) as means of initiation, booster (gun powder), and propellant (single base). The cartridge when suitably initiated burns and propels the seat out of the aircraft.

In fighter aircraft, there are eight types of cartridges in CSE set. Two of them are secondary cartridges. Cracks were observed on 18 Nos. of cartridges of similar type belonging to one lot of manufacture. The cracks were reported to have been observed after 3–4.5 years from the date of manufacture.

The cartridges were manufactured from Cu-Zn30 grade cartridge brass sheet of 0.9mm in 'HA'/'O' (quarter hard/annealed) condition conforming to specification IS:410. The processing involved blanking, first draw and second draw, trimming, and flattening. No stress-relieving operation was done during or after drawing. The cartridges were subjected to empty proof at +55°C and after filling with explosive at +45°C and -26°C as per laid down specification. The proof was satisfactory. Two each cracked and good cartridges were received by this laboratory for failure analysis.

10.4.2 Experimental Results

Visual Examination: The cracked cartridges in the as-received condition are shown in Fig. 10.4.1. The cracked portion of one of the cartridges is shown in Fig. 10.4.2. The wider part of the crack was found on the top U-bend of the cup. The crack was found to have propagated on either side of the U-bend. On opening the crack, the fracture surface showed tarnished regions at the U-bend and on the side wall (Fig. 10.4.3). Cup portion of



Fig. 10.4.1 Failed cartridges in the as-received condition.



Fig. 10.4.2 Side view of a sample of the failed cartridge showing the crack.

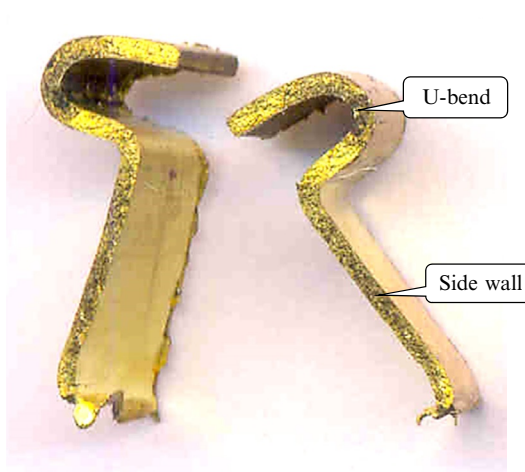


Fig. 10.4.3 Mating fracture surfaces of one of the cracks showing tarnished regions on the U-bend and side wall of the cup.

the cartridge is observed to have been secured to the lid at the end of the U-bend with a vulcanized rubber gasket.

Fractographic Examination: The fracture surface was cleaned ultrasonically in acetone and examined under scanning electron microscope (SEM). Near the inner surface of the U-bend the fracture surface showed discoloration (Fig. 10.4.4). At higher magnification the discolored region revealed mud cracking and corrosion debris (Fig. 10.4.5). EDAX of the corrosion debris showed the presence of sulfur in addition to the constituents of brass (Fig. 10.4.6). In general, the fracture surface appeared granular at low magnification. At higher magnification it showed an intergranular fracture with secondary intergranular cracks (Fig. 10.4.7). Some debris was found in pockets on the fracture surface (Fig. 10.4.8). EDAX revealed the debris to be rich in sulfur (Fig. 10.4.9).

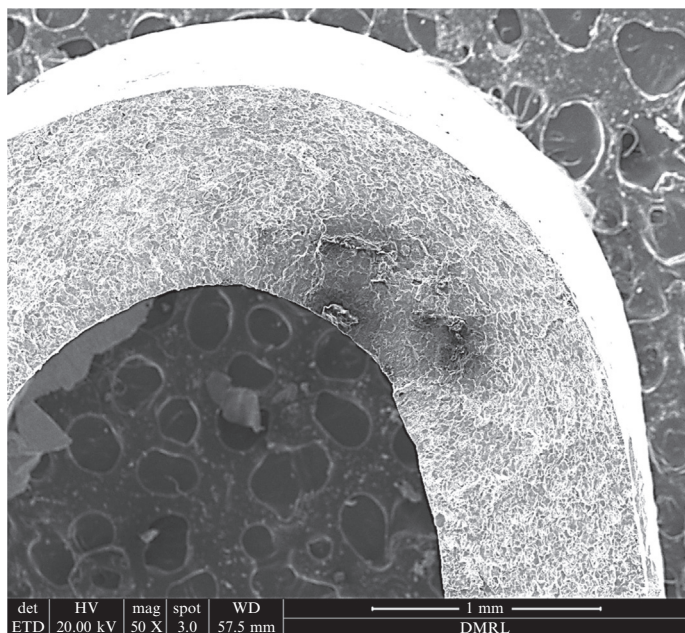


Fig. 10.4.4 SEM fractograph of the fracture surface showing tarnished and discolored region near the inner surface of the 'U'-bend.

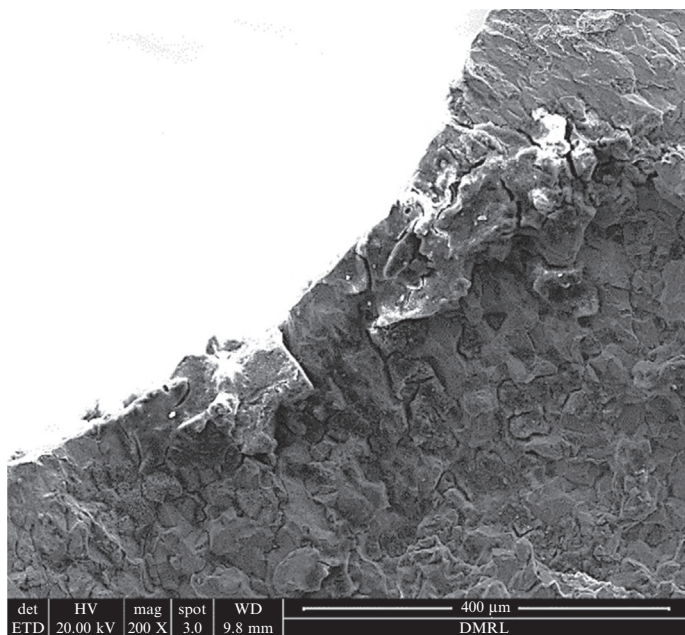


Fig. 10.4.5 SEM fractograph of the fracture surface showing corrosion debris and mud cracks in the discolored region.

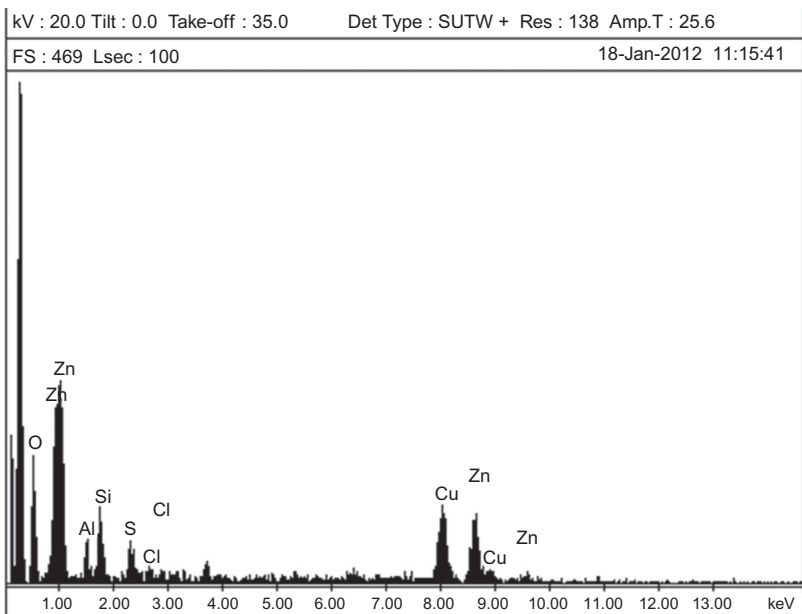


Fig. 10.4.6 EDAX of the corrosion debris showing presence of sulfur in addition to the elements composing the brass cartridge, the rubber gasket, and the filler compound.

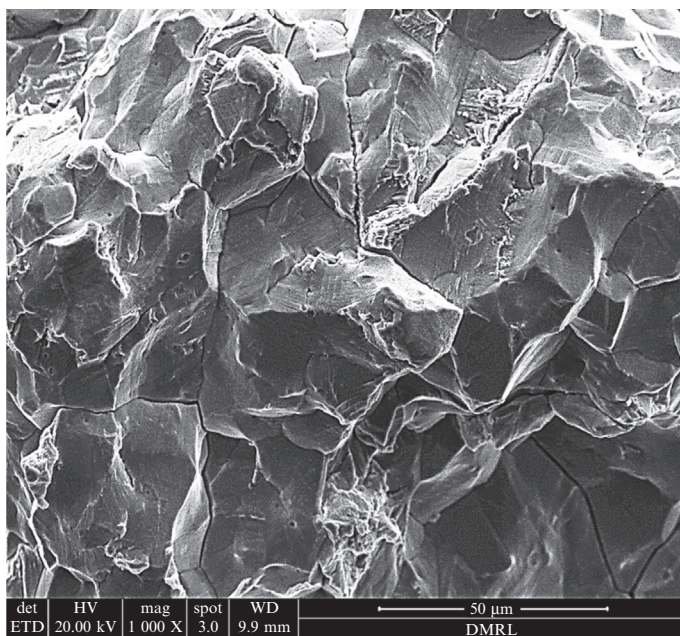


Fig. 10.4.7 SEM fractograph of the fracture surface showing an intergranular fracture with secondary cracks.

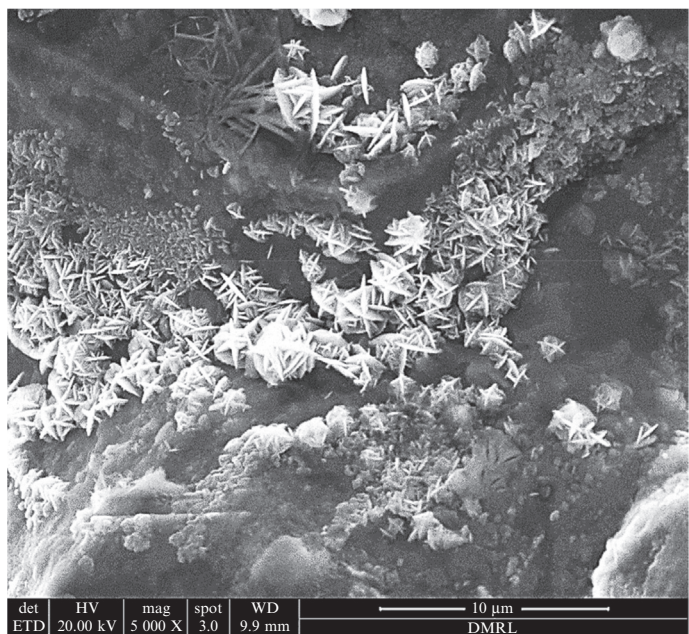


Fig. 10.4.8 SEM fractograph of the fracture surface showing pockets of debris at higher magnification.

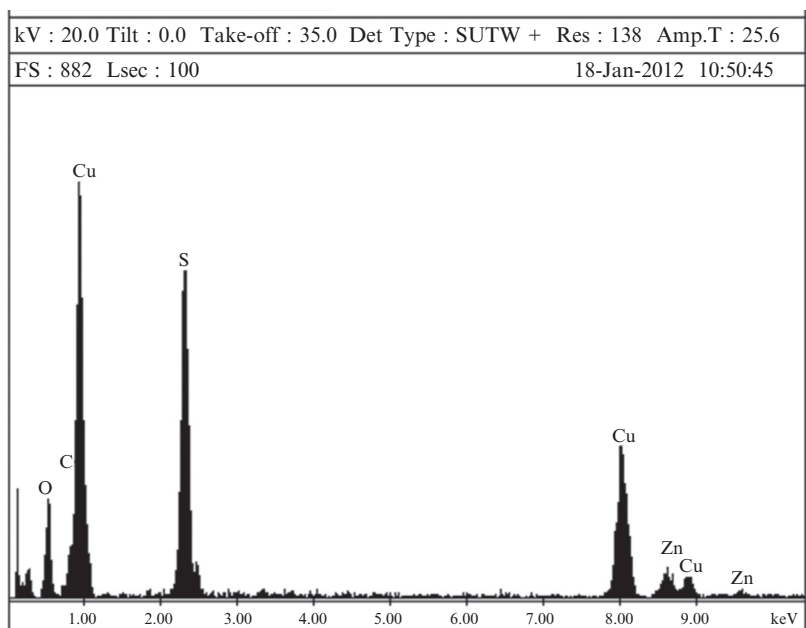


Fig. 10.4.9 EDAX of the debris showing presence of large amount of sulfur in addition to the elements constituting the brass cartridge.

Metallographic Examination: Sections from different regions of failed cartridge samples were prepared for metallographic examination. In the unetched condition the sample showed a large number of zinc sulfide inclusions (Fig. 10.4.10). On etching with acidified potassium dichromate solution section from the U-bend revealed corrosion near the inner surface (Fig. 10.4.11) and equiaxed alpha grains with some strain markings and zinc sulfide inclusions (Fig. 10.4.12). The section from the lid revealed larger equiaxed alpha grains with annealing twins and zinc sulfide inclusions



Fig. 10.4.10 Micrograph of a section from the cartridge sample showing strings of zinc sulfide inclusions. Unetched.

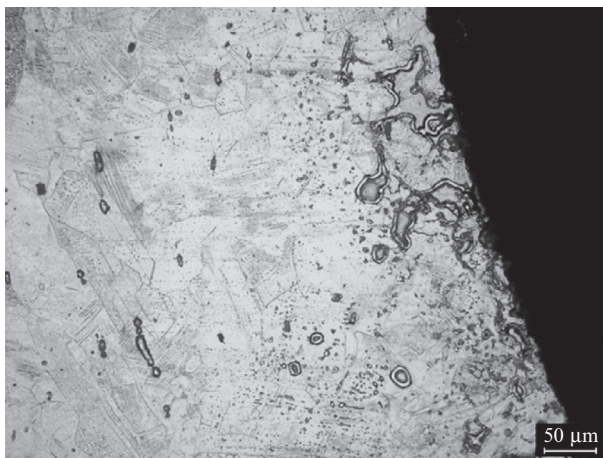


Fig. 10.4.11 Micrograph of a section showing corrosion near the inner surface of the U-bend. Etchant: Acidified potassium dichromate.



Fig. 10.4.12 Micrograph of a section from the U-bend of the cartridge sample showing equiaxed alpha grains with some strain markings and zinc sulfide inclusions. Etchant: Acidified potassium dichromate.



Fig. 10.4.13 Micrograph of a section from the lid of the cartridge sample showing equiaxed alpha grains with annealing twins and zinc sulfide inclusions. Etchant: Acidified potassium dichromate.

(Fig. 10.4.13). A section perpendicular to the fracture surface showed a number of secondary intergranular branching cracks (Fig. 10.4.14).

Hardness: The average Vickers hardness values measured at 5 kg load on different regions of cracked cartridge sample are as given here.



Fig. 10.4.14 Micrograph of a section perpendicular to the fracture surface of the cartridge sample showing a number of secondary intergranular branching cracks. Etchant: Acidified potassium dichromate.

- (a) Bottom of cup: 100 HV
- (b) Side of cup: 157 HV
- (c) U-bend of cup: 158 HV
- (d) Lid: 72 HV

Residual Stress Measurement: U-bend region of the cracked cartridge sample recorded a residual stress of + 35.0MPa.

10.4.3 Chemical Analysis (Wt.%)

Cu	S	Pb	Fe	Zn
69.0	—	Traces	—	Rem.

10.4.4 Discussion

Chemical analysis and metallographic examination suggested that the cartridges were made using brass nearly conforming to designation Cu-Zn30 of IS: 410-1977. Visual examination indicated that the cup was secured to the lid with a vulcanized rubber gasket at the top near U-bend. The crack in the cup was wider at U-bend and tapered on both sides, indicating that the crack initiated near the central region of the U-bend and propagated on both directions. Discoloration of the fracture surface on the U-bend and the side wall of cup suggested that the two regions were corroded. Corrosion due to

sulfur and its oxides could have initiated near the inner surface of the U-bend as evidenced by visual and fractographic examination/EDAX analysis. Presence of sulfur-rich particulate matter on the fracture surface may be due to embedment of explosive mixture that comprise of gun powder containing a substantial amount of sulfur in it. Sulfur in gun powder could have subsequently corroded the side wall of the cup after penetrating into the crack that propagated.

Large population of zinc sulfide inclusions and its segregation at the U-bend region of the cup could have aided in the initiation of micro cracks during cold working as evidenced by metallographic examination. A residual tensile stress of 35MPa was found present in the U-bend region of the cartridge sample. The residual stress could be the result of cold working. Although the brass sheet was supplied in the 'HA'/'O' (quarter hard/annealed) condition, the fabrication of cartridges involved cold working as evidenced by strain markings in the alpha grains and higher hardness of brass nearly to 'HB'/'HC' (half hard/hard) condition especially at side wall and U-bend regions of the cup. Inclusions, their segregation, and micro cracks are all regions of stress concentration which is found at the U-bend region of the cup. 70-30 cartridge brass is susceptible to stress corrosion cracking in the presence of sulfur and a tensile residual stress [1]. There were two sources from where sulfur could come in contact with the cup material. The first one is gun powder, which can ingress into the micro cracks present on the inner surface of the cup. The other one is vulcanized rubber used for gasket. Sulfur is the curator used in vulcanization of rubber. During prolonged storage rubber from gasket can leach into brass and corrode the metal [1]. Segregation of zinc sulfide inclusions has aided in the formation of micro cracks during cold working which in turn have aided in corrosion of the metal.

The combined action of

- (a) segregation of zinc sulfide inclusions and micro cracking at the U-bend during cold working of the cup,
- (b) leaching of rubber from gasket and getting lodged into the micro cracks,
- (c) ingress of gun powder into micro cracks,
- (d) corrosion of brass due to the presence of sulfur and its oxides under suitable conditions,
- (e) presence of residual tensile stress and stress concentration at the U-bend due to cold working has been responsible for stress corrosion cracking of the cartridges.

The stress corrosion cracking is confirmed by the presence of secondary intergranular branching cracks and an intergranular fracture as evidence in metallographic and fractographic examinations.

10.4.5 Conclusion

The cartridges of the seat ejection system failed due to stress corrosion cracking.

10.4.6 Recommendations

- (i) Use cartridge brass free of zinc sulfide inclusions.
- (ii) Stress relieve the brass after cold working.
- (iii) Use better quality of rubber gasket that has longer life.

10.4.7 References

- [1] ASM Handbook, Vol-11 'Failure Analysis and Prevention', 2002 edition, ASM International, Materials Park, Ohio, pp. 1738–1820.



Wear

Wear of metal parts in engineering systems affect both performance as well as cost considerably. There are different types of wear. More prominent among them are abrasive wear, erosive wear, and adhesive wear.

Abrasive wear: When hard particles move against a solid surface the solid suffers an abrasive wear. Abrasion is manifested by the formation of surface grooves and the damage suffered could be due to cutting, plowing, or chipping of the solid. The mechanism of material removal due to abrasion depends on the mechanical properties of the solid surface and the geometry of the abrading particles. In general abrasion resistance of a ductile single phase material depends on its hardness, whereas for a brittle material its fracture characteristic determines the abrasion behavior. In alloys designed to resist abrasion size, shape and dispersion of a hard phase within the matrix determine their abrasion resistance.

Erosive wear: When particles of a solid, liquid, or gas impact on a solid metal surface at high velocity the solid metal surface suffers an erosive wear. The damage caused to the solid metal surface depends on the nature, velocity, and angle of impingement of the particles. The damage suffered due to erosion is observed as micro-cutting, micro-plowing, micro-cracking, surface and subsurface fatigue cracks, extrusion of material at the exit end, etc. Erosion rate is one of the important properties studied against influencing operating parameters viz. exposure time, impact angle, particle size, and impact velocity. Usually erosion rate is defined as mass loss of a material divided by the mass of impacting particles.

Hardness, work hardening, and capability of deformation are important properties of the target metal for its resistance to erosive wear. However, there is no one-to-one relationship between erosive wear to a material property since the operating parameters influence particle erosion. Moreover microstructure also has an effect on the erosion behavior of a metal.

Adhesive wear: When metallic surfaces slide against each other it results in wear of the weaker of the mating surfaces. The process of such a wear is called adhesive wear. Strong interfacial bonds may occur at deformed surface

asperities, with mechanical degradation arising from subsequent shear failure in the weaker of the mating surfaces. Subsurface crack nucleation and growth by fatigue causing shearing of surface asperities is another mechanism of adhesive wear.

Reduction of the strength of interfacial bonds is essential for preventing adhesive wear. This can be achieved by separating the rubbing surfaces by a lubricant. Sometimes the effect of lubrication may change under unfavorable condition of operation of machines, for instance during run-in periods, start-up and shutdown. In such situations of boundary friction, the surfaces may be separated by a very thin film of lubricant. Contact stresses and heating can destroy easily this thin film and cause seizure of the contacting surfaces. Bearing seizure is often the cause of bearing failure due to ineffective lubrication of the system. Hardened steels are substantially less liable to seizure than normalized or annealed ones. For that reason, hardened steels and those hardened by surface hardening treatments are the principal materials to make mating friction pairs.

In order to overcome friction among contacting surfaces and avoid premature wear of the parts of certain engineering devices viz. bearings, gears, axles, shafts, pistons and cylinders, etc. are minimized by suitably lubricating them. Lubricants are solid, semisolid, or liquid substances having the required viscosity in order to oppose the forces that facilitate contacts between surfaces. Lubricants also have the property of dissipating frictional heat that is generated by sliding parts of the device.

Effective lubrication is the prime requirement in the smooth functioning of devices wherein moving parts are in contact. Lubricating oils, grease, and solid lubricants like molybdenum disulfide are commonly used for lubrication. Lubricants become ineffective due to chemical decomposition, contamination, and degradation due to heat or a combination of one or more of these processes. Failure of lubrication can cause dry running conditions of moving parts, resulting in frictional overheating and subsequent failure of the device. Contamination of lubricant by external environment results in the formation of substances that can cause corrosion or wear of the parts.

Failures by ineffective lubrication can be prevented by using (a) appropriate lubricant that has longer life, (b) using additives with lubricant that improve its properties resulting in better efficiency and life of the lubricant, (c) improving the mechanical design of the device, and (d) incorporating suitable filters into the device that prevent contamination of the lubricant.



11.1 CASE STUDY: FAILURE OF INTERROTOR BEARING OF AN AEROENGINE

11.1.1 Background

The rotors of an aeroengine were found to be jammed after landing with the problem of in-flight high vibration warning. On strip examination of the engine, the interrotor bearing was found to be damaged. The total operating time of the engine was reported as 677 h since new. From the available information it was found that 10W-5Cr-1V-1C steel was used for the bearing races and rollers, while 11Al-4Fe-Cu bronze was used for the cage of the bearing assembly.

11.1.2 Experimental Results

Visual Examination: The failed bearing in the as-received condition is shown in Fig. 11.1.1. The inner surface of the outer race showed pitting and grooving of the surface on one side (Fig. 11.1.2). The outer surface of the inner race showed score marks (Fig. 11.1.3). The rollers were badly worn out and a few showed flattened surfaces (Fig. 11.1.4). The outer and inner surfaces of the cage showed severe wear and breakage (Fig. 11.1.5).



Fig. 11.1.1 Failed bearing showing the outer race, the inner race, the rollers, and the cage in the as-received condition.



Fig. 11.1.2 Inner surface of the outer race showing the damage concentrated to one side

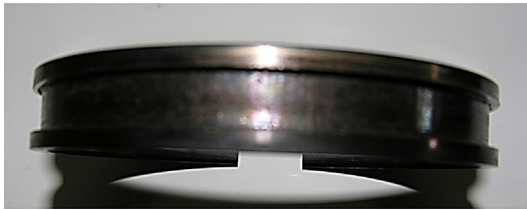


Fig. 11.1.3 Outer surface of the inner race showing score marks.

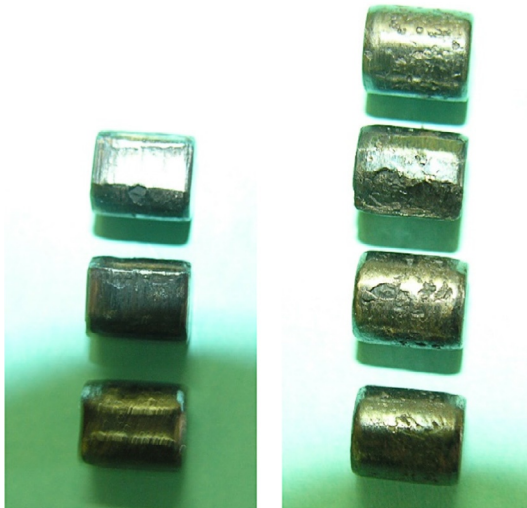


Fig. 11.1.4 Rollers showing severe wear and flattening of the surfaces.

Fractographic Examination: The fracture surface of the cage was cleaned with mild oxide remover and further cleaned in acetone ultrasonically and examined under scanning electron microscope (SEM). The fracture surface revealed a quasi-cleavage fracture ([Fig. 11.1.6](#)).



Fig. 11.1.5 Outer and inner surfaces of the cage showing surface wear and breakage.

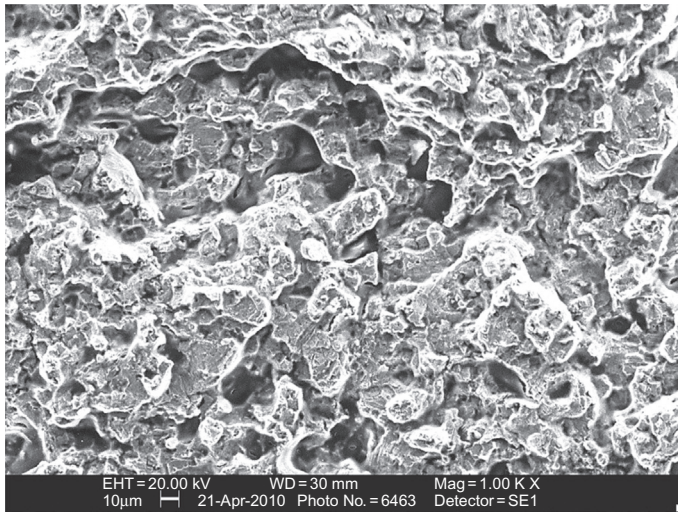


Fig. 11.1.6 SEM fractograph of the fracture surface of the cage showing quasi-cleavage fracture feature.

Metallographic Examination: Samples from outer race, inner race, rollers, and cage were prepared for metallographic examinations.

Outer Race: In the unetched condition the inner surface of the outer race showed a number of surface and subsurface cracks parallel to the surface (Fig. 11.1.7). The inner surface of the outer race also showed surface spalling

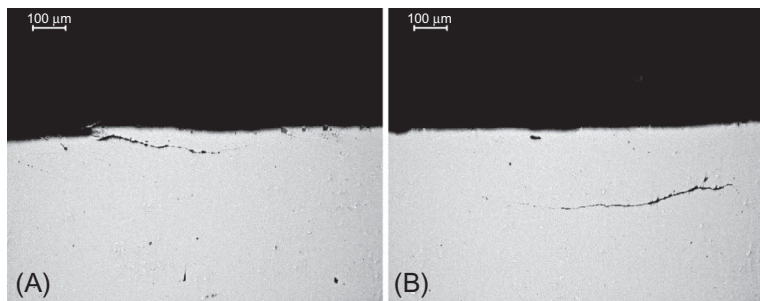


Fig. 11.1.7 Micrograph of the inner surface of the outer race showing crack near the (A) surface and (B) subsurface crack almost parallel to the surface. Unetched.

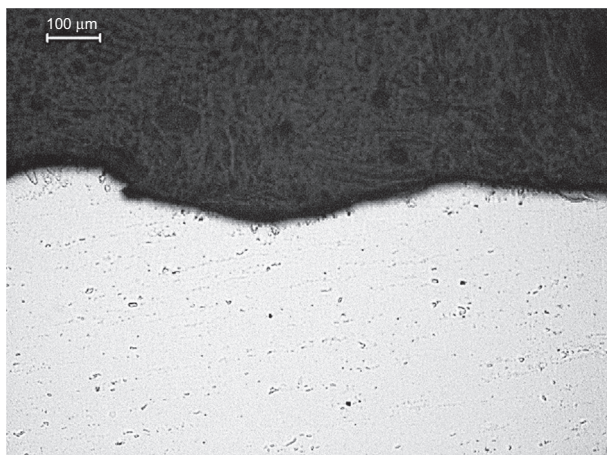


Fig. 11.1.8 Micrograph of the inner surface of the outer race showing spalling and removal of material from the surface. Unetched.

(Fig. 11.1.8). On etching with 2% nital, the sample revealed undissolved (coarse) and reprecipitated (fine) carbides in a matrix of tempered martensite (Figs. 11.1.9 and 11.1.10).

Inner Race: In the unetched condition, the outer surface of the inner race showed cracking and spalling at the corners of the groove (Fig. 11.1.11). The flat surface at the center showed micro-pitting and cracks (Fig. 11.1.12). On etching with 2% nital the sample revealed undissolved (coarse) and reprecipitated (fine) carbides in a matrix of tempered martensite (Fig. 11.1.13).

Rollers: In the unetched condition, longitudinal section of the rollers showed surface wear, cracks, and removal of material (Fig. 11.1.14). On

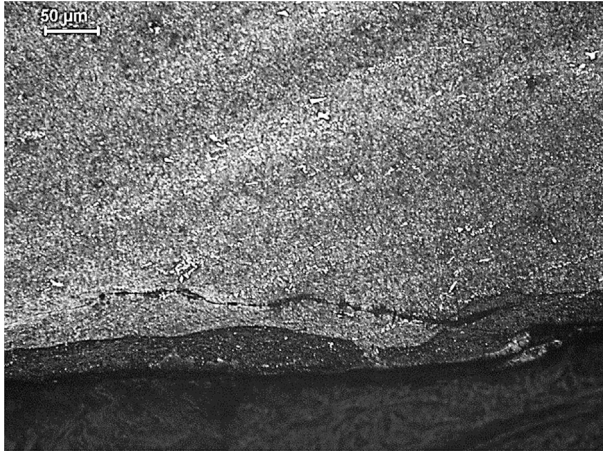


Fig. 11.1.9 Micrograph of the inner surface of the outer race showing cracking and spalling near the surface. Etchant: 2% nital.

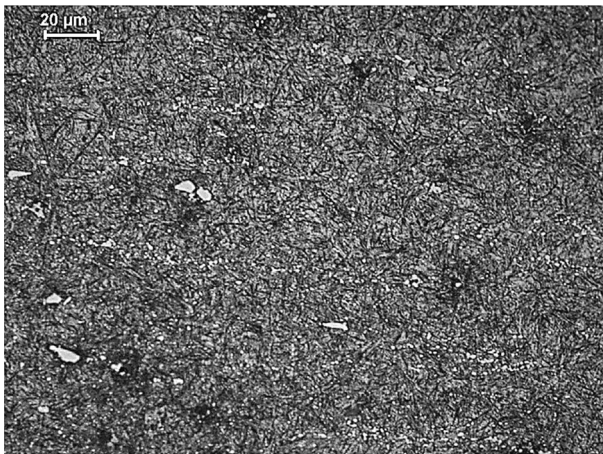


Fig. 11.1.10 Micrograph of the outer race showing undissolved (coarse) and reprecipitated (fine) carbides in a matrix of tempered martensite. Etchant: 2% nital.

etching with 2% nital, the sample revealed carbides in a matrix of tempered martensite (Fig. 11.1.15).

Cage: In the unetched condition the cage sample showed surface wear, cracks, and removal of material (Fig. 11.1.16). On etching with acidified potassium dichromate solution, the sample revealed alpha grains in a matrix of transformed eutectoid beta (Fig. 11.1.17).

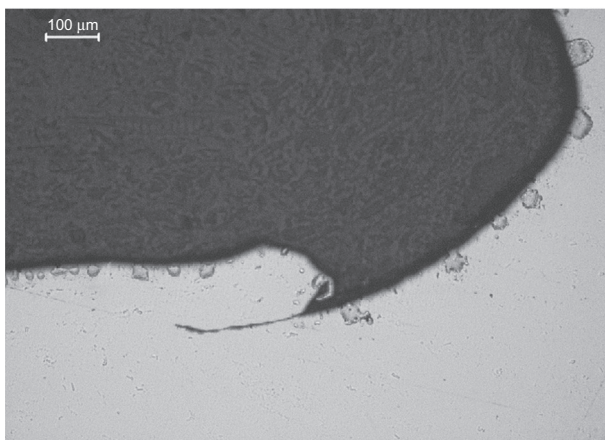
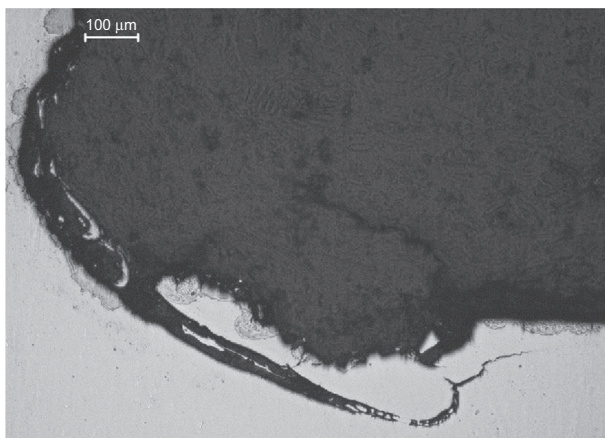


Fig. 11.1.11 Micrograph of the outer surface of the inner race showing cracking and spalling of the material from the corners of the grooves. Unetched.

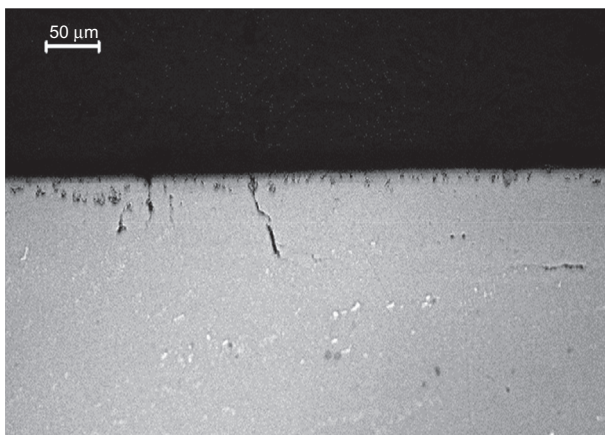


Fig. 11.1.12 Micrograph of the outer surface of the inner race showing micro-pitting and cracking near the surface. Unetched.

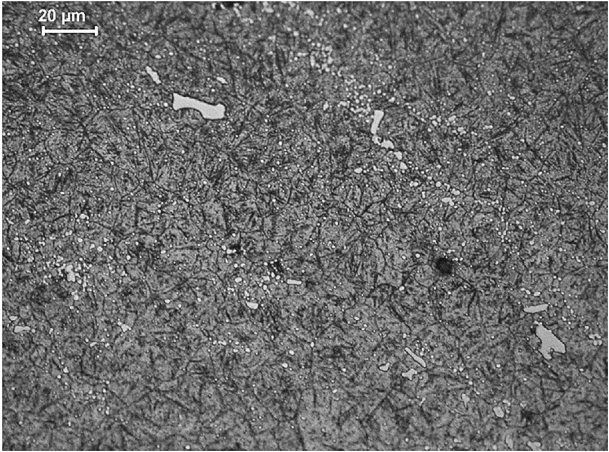


Fig. 11.1.13 Micrograph of the inner race showing undissolved (coarse) and reprecipitated (fine) carbides in a matrix of tempered martensite. Etchant: 2% nital.

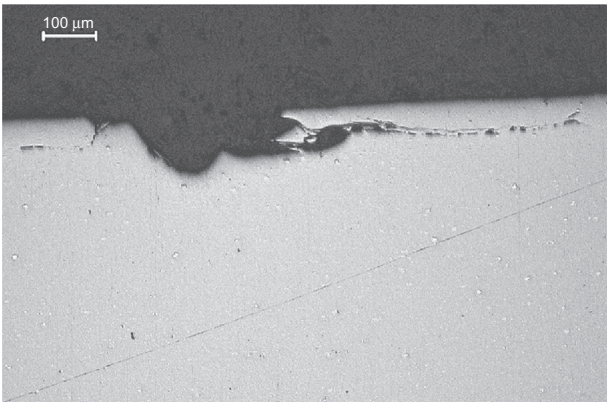


Fig. 11.1.14 Micrograph of horizontal section of a roller showing surface erosion and removal of material. Unetched.

Hardness: Average Vickers hardness measured at 30kg load was as given here.

Outer race	851 HV
Inner race	868 HV
Roller	672 HV
Cage	276 HV

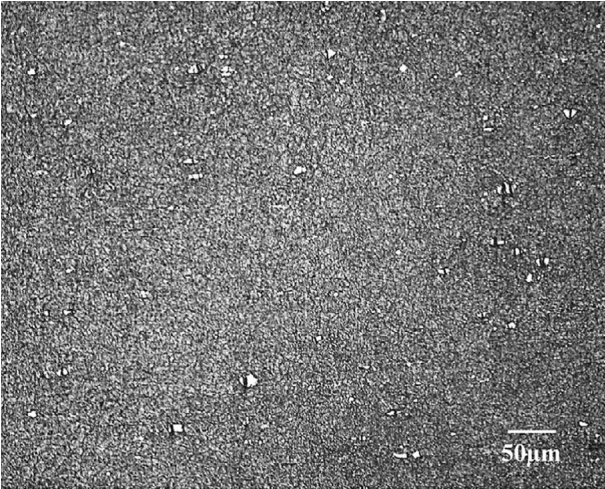


Fig. 11.1.15 Micrograph of the roller sample showing undissolved (coarse) and reprecipitated (fine) carbides in a matrix of tempered martensite. Etchant: 2% nital.



Fig. 11.1.16 Micrograph of the cage sample showing surface erosion and removal of material. Unetched.

11.1.3 Chemical Analysis (Wt.%)

<i>Outer Race, Inner Race, and Roller</i>						
C	Si	Mn	Cr	W	V	Fe
0.74	0.30	0.14	5.01	10.2	1.08	Balance
<i>Cage</i>						
Al	Mn		Fe		Cu	
11.0	2.2		3.5		Balance	

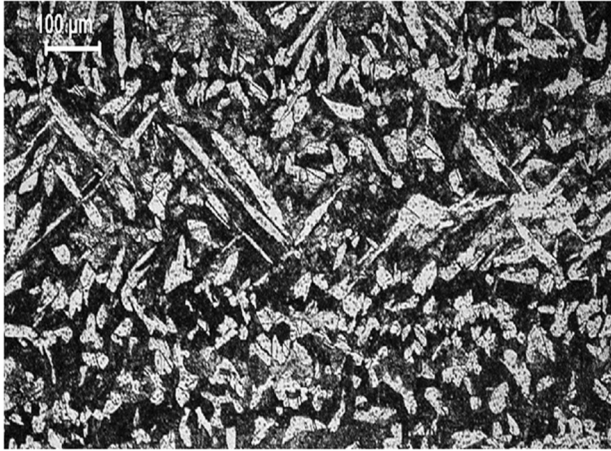


Fig. 11.1.17 Micrograph of the cage sample showing alpha grains in a matrix of transformed eutectoid beta. Etchant: Acidified potassium dichromate.

11.1.4 Discussion

Chemical composition of the interrotor bearing (i.e. outer race, inner race, and rollers) nearly conformed to the specified grade of steel. The cage was also found to be made nearly of the specified grade of aluminum bronze. Microstructure of the outer race, inner race, and rollers showed hardened and tempered condition. However, the hardness of the rollers was much less than that usually found in such bearings. This may be due to tempering of the rollers caused due to frictional heating as a result of rubbing during restrictions in their motion.

The outer race showed pitting and grooving on the inner surface a little off-center. This happens when the radial load exceeds the axial load and when the bearing is overloaded. The outer race also revealed surface and subsurface cracks and spalling of the inner surface, which is indicative of rolling contact fatigue.

Fatigue cracks are initiated at or very near the contact surfaces of bearing components as a result of stress concentrations. High stresses in localized regions are caused due to improper contact configurations, misalignment, or distortion. Eventually, such situation can lead to surface deterioration, crack initiation, and spalling. Alternating shear stresses reach maximum values at a distance below the surface initiating subsurface cracks. Subsurface-initiated fatigue cracks propagate parallel to the surface and eventually result in dish-shaped spall.

The outer surface of the inner race revealed cracks and spalling at the corners of grooves. Small micro-pitting and cracking were also observed at the polished contact surface. These are again evidences of rolling contact fatigue.

Surface wear of rollers by pitting has resulted due to rattling movement of the rollers while sliding/skidding over damaged raceways causing friction and heating. The roller surfaces were deformed due to rubbing action at high temperature as evidenced by their flat surfaces and lower hardness.

The investigation did not show any evidence of oil starvation or contamination of oil. Skidding/sliding could have taken place instead of rolling in the roller bearing due to misalignment and improper contact as evidenced by damages off the center seen on the inner surface of the outer race and the surface damages on the rollers. There does not seem to be any material-related deficiency or metallurgical abnormality (microstructure) of the inter-rotor bearing material.

11.1.5 Conclusion

1. The interrotor bearing had failed due to rolling contact fatigue.
2. The failure does not seem to be due to oil starvation or oil contamination.
3. Skidding/sliding could have taken place instead of rolling due to misalignment and improper contact.



11.2 CASE STUDY: FAILURE OF LOW TEMPERATURE SUPER HEATER TUBE OF A THERMAL POWER STATION

11.2.1 Background

Low temperature super heater (LTSH) tube pertaining to the boiler of 120 MW thermal power station failed prematurely in service. The tube was surrounded by an atmosphere of flue gas at 600°C. Steam at a pressure of 144 kg/cm² and temperature of 450°C flowed through the tube.

11.2.2 Experimental Results

Visual Examination: The LTSH tube in the as-received condition is shown in Fig. 11.2.1. The tube was of 43 mm O.D and 5 mm thickness at a location away from failure that ought to be the original dimensions. The failure was found to have occurred at a bend. At the failed portion over a length of 150 mm and 30 mm along the circumference severe blistering and erosion of the outer surface was seen (Fig. 11.2.2). The thickness of the tube near the bend had reduced to 2 mm and it had thinned down to <0.5 mm at the failed zone (Fig. 11.2.3). No definite origin of the failure could be

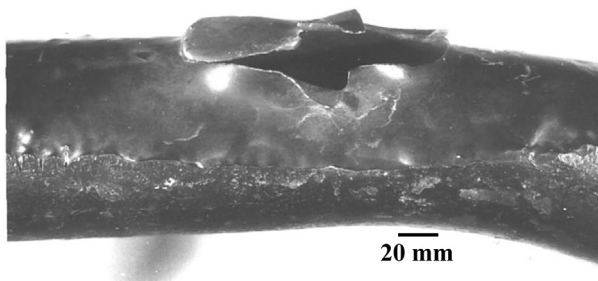


Fig. 11.2.1 The failed LTSH tube in the as-received condition.

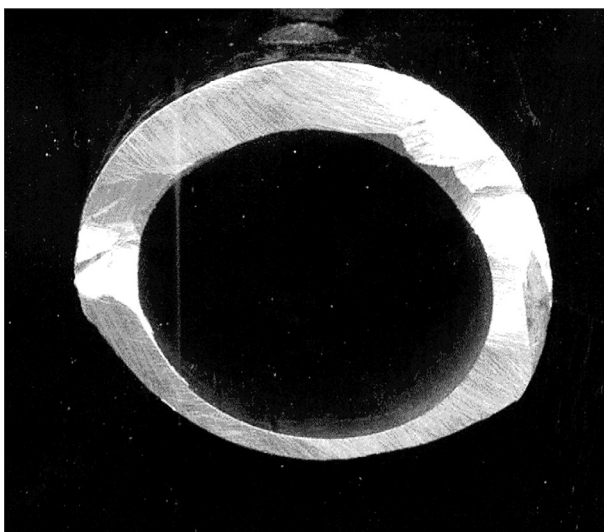


Fig. 11.2.2 Cross section near the failed portion of the tube showing severe surface erosion evidenced by reduction in thickness.



Fig. 11.2.3 Thinning of the failed region seen on the tube.

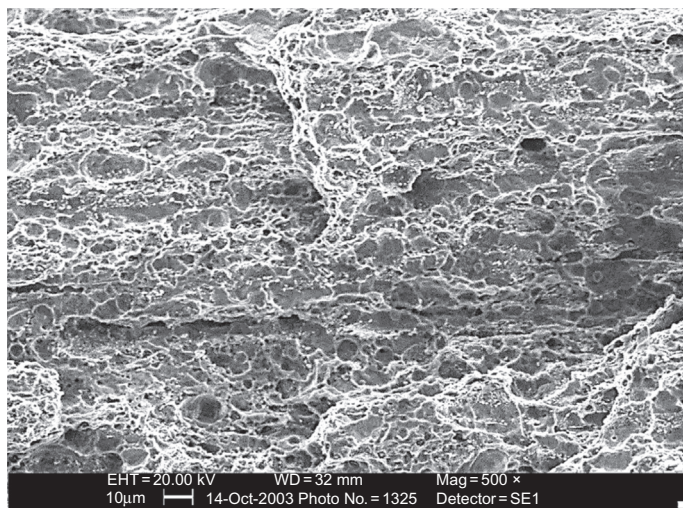


Fig. 11.2.4 SEM fractograph of the fracture surface showing slightly elongated dimples.

located on the fracture surface, which showed fibrous fracture under stereomicroscope.

Fractographic Examination: On examining the fracture surface under SEM, the specimens revealed slightly elongated dimples (Fig. 11.2.4).

Metallographic Examination: Sections of the failed tube both near and away from failure were prepared for metallography. On etching with 2% nital, the specimen near the failure showed elongated grains of ferrite and pearlite (Fig. 11.2.5). The specimen away from failure revealed equiaxed grains of ferrite and pearlite (Fig. 11.2.6).

Hardness: Average Vickers hardness measured on the sample at 10 kg load was 207 HV near failed region and 144 HV away from it.

11.2.3 Chemical Analysis (Wt.%)

C	Si	Mn	S	P	Fe
0.27	0.72	0.62	0.011	0.02	Rem.

11.2.4 Discussion

Chemical composition suggested that the LTSH tube was made of a plain carbon steel nearly conforming to BS 970-1955, En5 grade. Microstructural examination and hardness of a sample away from the failed region indicated

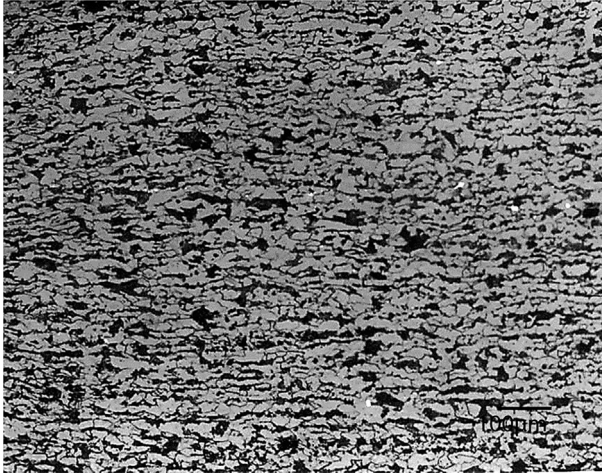


Fig. 11.2.5 Micrograph of a section of the tube showing elongated grains of ferrite and pearlite near the failure. Etchant: 2% nital.

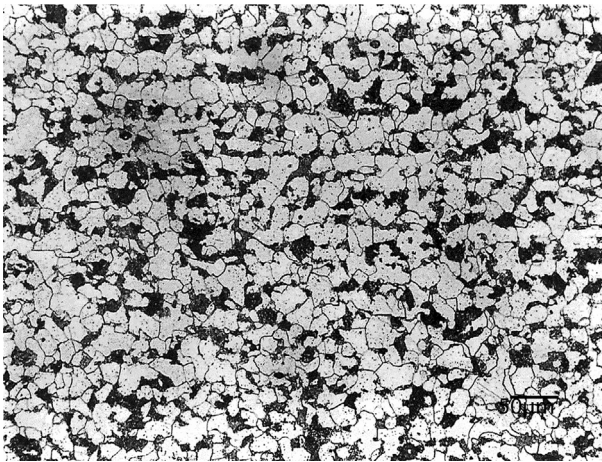


Fig. 11.2.6 Micrograph of a section of the tube showing equiaxed grains of ferrite and pearlite away from the failure. Etchant: 2% nital.

that the tube was used in annealed condition. The tube showed severe blistering and erosion on the outer surface leading to considerable thinning of the tube section from 5 mm to <2 mm. The surface erosion could have been caused by flew gas impingement on the bend portion of the tube as evidenced by maximum material removal in that region. Blistering and material removal coupled with discoloration in the area suggested hot

erosion to be the primary cause of failure. The erosion had caused the thinning of the tube to such an extent that it could no longer withstand the applied load at the prevailing temperature. High temperature of flue gas (600°C) on the outer surface and super-heated steam (450°C) on the inner surface caused lowering of the yield strength of the steel. However, metallography confirmed that the overheating of the tube has not taken place beyond AC_1 as evidenced by cold worked structure and higher hardness of the failed region as compared to equiaxed structure and lower hardness of an area away from failure. Lower yield strength due to prevailing temperature conditions coupled with thinning of the tube section due to hot erosion could no longer permit the LTSH tube to withstand the applied load causing overload failure of the tube as evidenced by dimpled rupture.

11.2.5 Conclusion

The LTSH tube had failed primarily due to hot erosion of the outer surface probably due to impingement of hot flew gases focused at the bend. The thinning of the tube due to hot erosion and low yield strength caused by high temperature resulted in final failure of the tube by overload.



11.3 CASE STUDY: FAILURE OF MAIN LANDING GEAR ASSEMBLY OF AN AIRCRAFT

11.3.1 Background

During routine sortie of an aircraft the port main wheel came out from the main landing gear assembly. Due to the dislodgement of the wheel the brake unit had sustained the entire landing load on the port side.

The components of the main landing gear assembly viz. the wheel hub, the axle (oleo), and the brake unit were received for failure investigation. The assembly had completed 426 h of service after last overhaul, while its total technical life was not mentioned. After last lubrication, the assembly had completed 31 landings and 7 h flying.

11.3.2 Experimental Results

Visual Examination: The wheel hub did not show any sign of failure. The brake unit was badly rubbed and worn out. The axle was found broken at the threaded portion (Fig. 11.3.1). Close-up of the broken portion of the axle is shown in Fig. 11.3.2. The mating fracture portion of the axle was inside the axle nut. The outer bearing in the as-received condition is

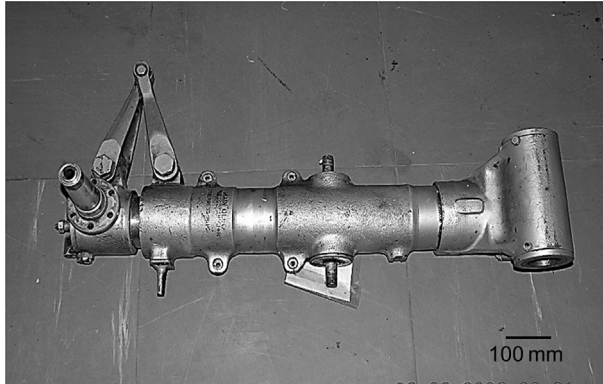


Fig. 11.3.1 Oleo in the as-received condition showing the broken axle.

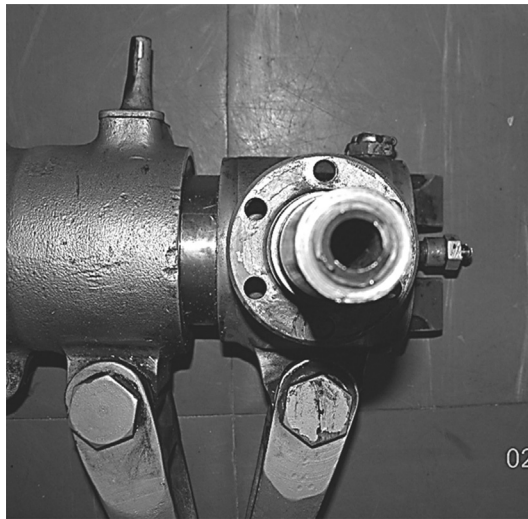


Fig. 11.3.2 Close-up of the fractured portion of the axle.

shown in [Fig. 11.3.3](#). The outer surface of the inner race of the bearing showed dent marks on at least half of the circumference at locations corresponding to the rollers ([Fig. 11.3.4](#)). The inner surface revealed scoring marks and bluing. Only three rollers were found which were worn out badly and showed evidences of bluing. The cage was badly rubbed, deformed, and broken. The split pin which locks the axle nut to the axle at the thread was found missing. Since the axle was broken and the bearing showed evidences of failure, the two components were taken up for further examination.



Fig. 11.3.3 The outer bearing in the as-received condition showing the inner race, the cage, and the rollers.



Fig. 11.3.4 The outer surface of the inner race showing dent marks corresponding to the locations of the rollers.

The fracture surface of the axle showed a 45° shear with one half showing a cup portion and the other half a cone portion. The junction of the two portions showed a twist. The fracture had originated from the first thread root.

Fractographic Examination: The portion of the axle remaining within the axle nut was unscrewed, cleaned with acetone ultrasonically, and the fracture surface was examined under scanning electron microscope. The fracture surface showed a fibrous fracture originating from the first thread root (Fig. 11.3.5). The fracture surface revealed equiaxed dimples all over it (Fig. 11.3.6).

Metallographic Examination: Sections from the inner race, roller, and cage of the bearing and the axle from the thread region were mounted and

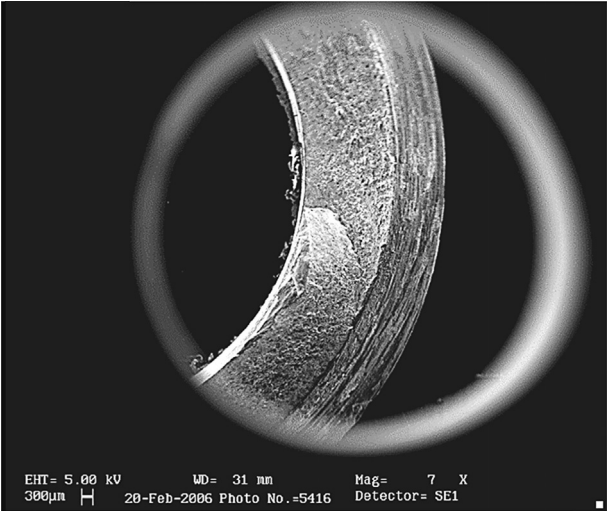


Fig. 11.3.5 SEM fractograph of the fracture surface of the axle showing a fibrous fracture.

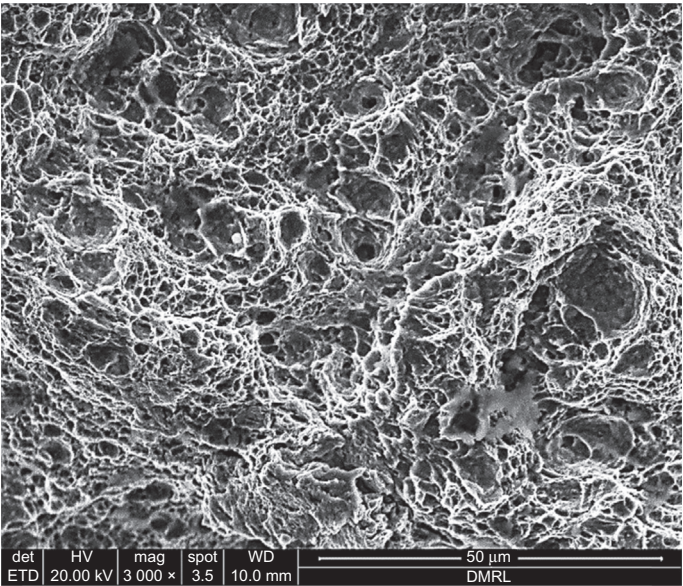


Fig. 11.3.6 SEM fractograph of the fracture surface of the axle showing equiaxed dimpled rupture.

prepared for metallography. In the unetched condition, the inner race of the bearing showed zig-zag crack from both outer surface as well as inner surface (Fig. 11.3.7). On etching with 2% nital, the sample showed severe deformation and crack at the elbow (Fig. 11.3.8). The inner as well as outer surfaces showed white etching layer, an evidence of martensitic transformation due to overheating (Figs.11.3.9 and 11.3.10). The sample revealed a tempered martensite microstructure with prominent acicularity (Fig. 11.3.11).

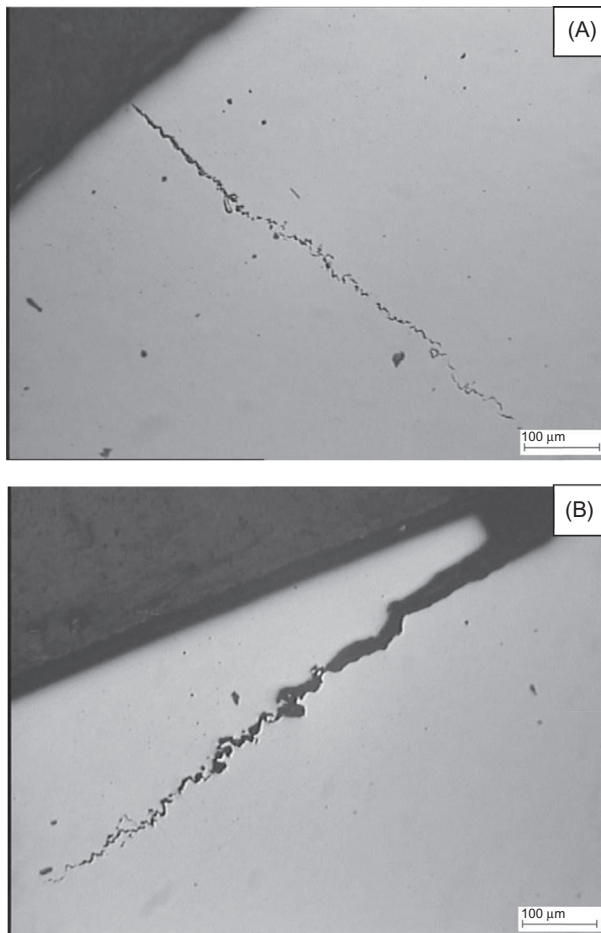


Fig. 11.3.7 Micrograph of transverse section of the inner race sample showing cracks at the (A) inner and (B) the outer surfaces. Unetched

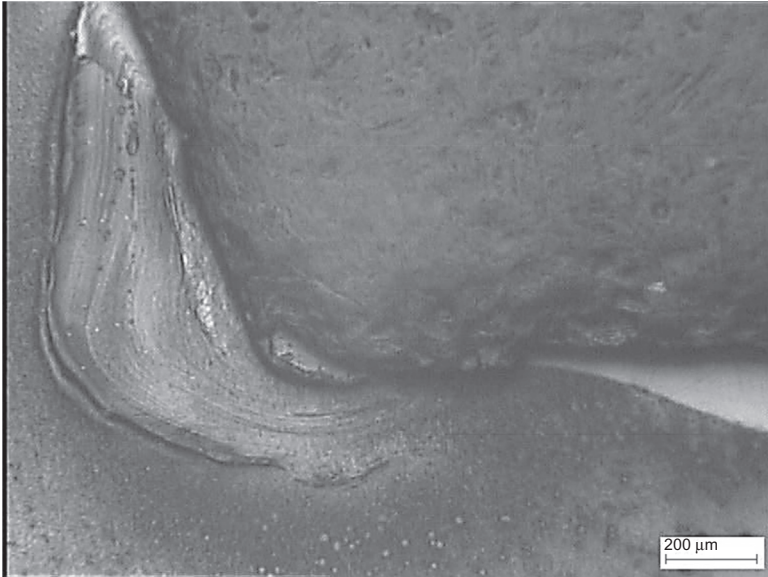


Fig. 11.3.8 Micrograph of the transverse section of the inner race showing severe deformation and cracking at the elbow. Etchant: 2 % Nital.

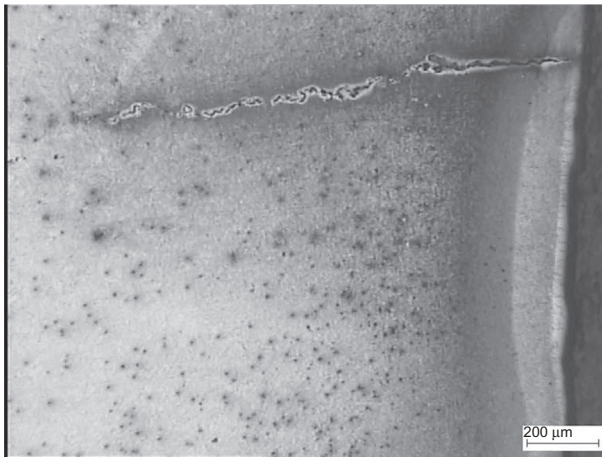


Fig. 11.3.9 Micrograph of the transverse section of the inner race showing white etching layer of untempered martensite and crack on the inner surface. Etchant: 2 % Nital.

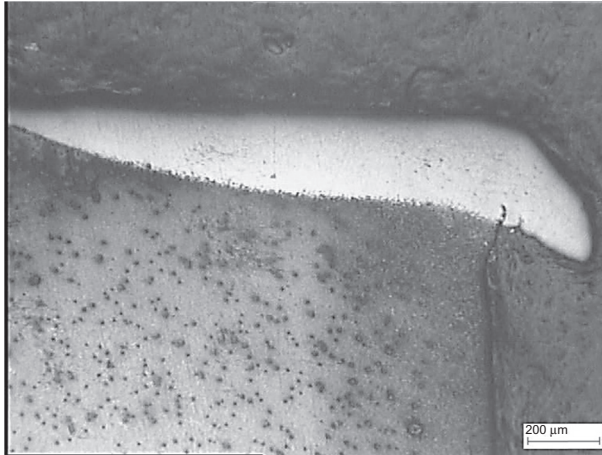


Fig. 11.3.10 Micrograph of the transverse section of the inner race showing white etching layer of untempered martensite on the outer surface. Etchant: 2 % Nital.

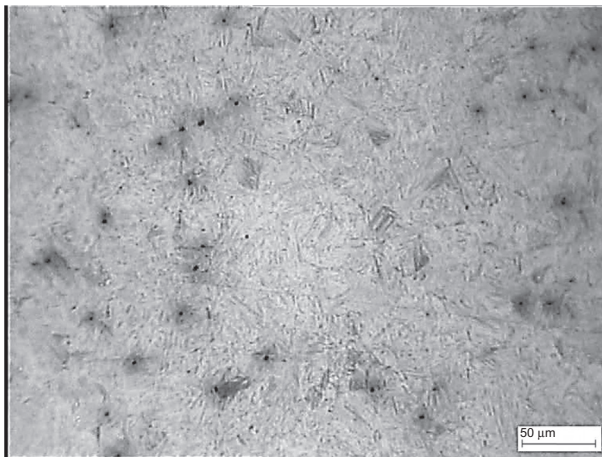


Fig. 11.3.11 Micrograph of the transverse section of the inner race showing a tempered martensite microstructure. Etchant: 2% Nital.

In the unetched condition the roller sample showed deformation on the surface. On etching with 2% nital the sample revealed white etching layer of untempered martensite on the surface near the deformation ([Fig. 11.3.12](#)) and tempered martensite in the core ([Fig. 11.3.13](#)). The cage sample showed a deformed ferrite-pearlite microstructure ([Fig. 11.3.14](#)).

The section of the axle showed some paint like coating on the surface. On etching with 2% nital, the sample showed a heavily banded microstructure. The sample showed a tempered martensite structure at higher magnification ([Fig. 11.3.15](#)).

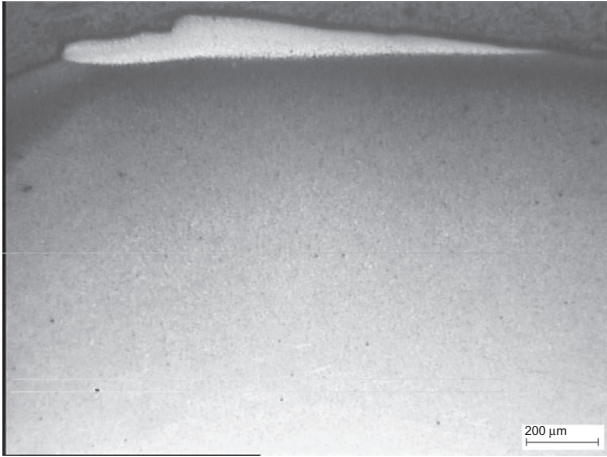


Fig. 11.3.12 Micrograph of the transverse section of the roller sample showing white etching layer of untempered martensite on the surface. Etchant: 2% Nital.

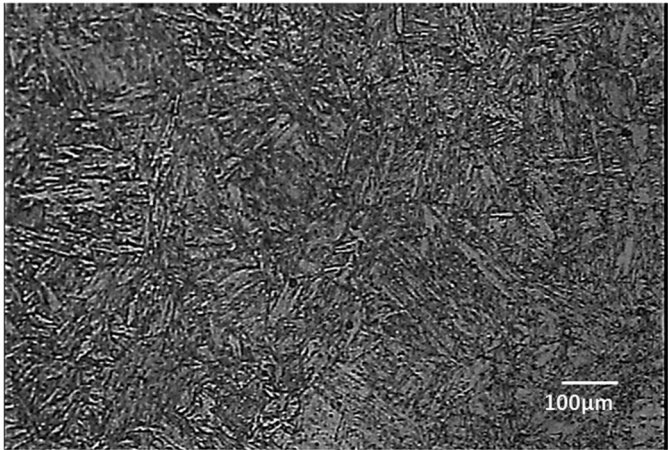


Fig. 11.3.13 Micrograph of the transverse section of the roller sample showing a tempered martensite microstructure. Etchant: 2% Nital.

Hardness:

Sample	Hardness (HV)
Bearing race	622
Roller	427
Cage	181
Axle	422

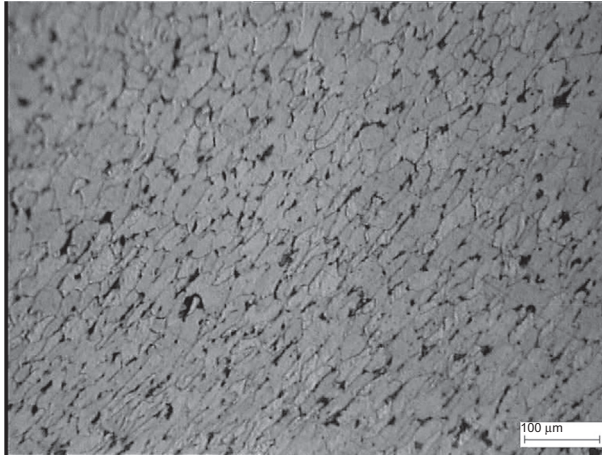


Fig. 11.3.14 Micrograph of the transverse section of the cage sample showing deformed grains of ferrite and pearlite. Etchant: 2% Nital.

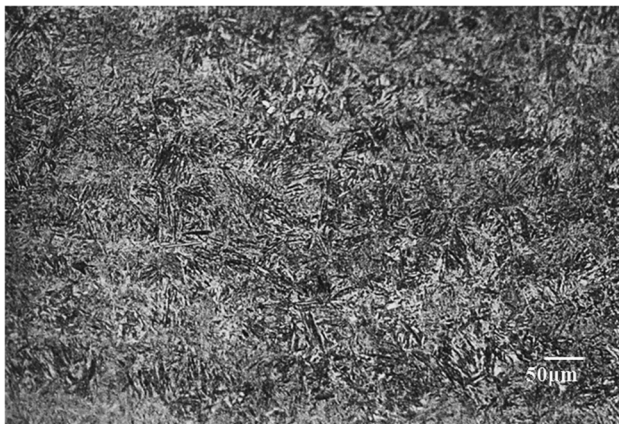


Fig. 11.3.15 Micrograph of the longitudinal section of the axle sample showing a tempered martensite microstructure. Etchant: 2% Nital.

11.3.3 Chemical Analysis (Wt.%)

Axle:

C	Cr	Mn	Si	Ni	Mo	Fe
0.33	1.00	0.5	0.25	0.25	0.19	Base

Bearing Race, Rollers, and Cage:

C	Cr	Mn	Si	Ni	Fe
1.03	1.46	0.45	0.37	0.24	Base

11.3.4 Discussion

The investigation indicated that the two major components that had caused failure of the main landing gear of the aircraft were the outer bearing and the axle.

Chemical analysis of the axle suggested that it was made of a low alloy steel conforming to SAE 4130. Microstructure and hardness indicated that it was used in a hardened and tempered condition. Fibrous fracture and dimpled rupture pointed to the fact that the axle had suffered an overload failure.

Chemical analysis of the outer bearing components suggested that they were made of a low alloy steel conforming to SAE 52100. Microstructure indicated that the races and rollers were used in hardened and tempered condition, while the cage was used in annealed condition. Microstructural examination revealed evidences of frictional overheating of both inner race as well as the rollers. Frictional overheating has resulted in heating the parts to austenitizing temperature of the bearing steel and subsequent cooling by the mass of the components. This has resulted in the formation of untempered martensite as evidenced by white etching layer on the surface of both inner race and the rollers. Denting of the outer surface of the inner race by the rollers was due to the application of stress at high temperature. The rollers have also revealed evidences of deformation and low hardness which again confirm overheating and overload at high temperature conditions. The cage is generally used in spheroidized annealed condition. However, the deformed ferrite and pearlite structure in the cage sample is an indication of overheating or deformation just below the AC_1 temperature of the steel used.

Frictional overheating could be the result of dry running conditions of the bearing. This has led to bearing seizure causing score marks and micro-cracks on the inner and outer surfaces of the inner race. The circumstances prevailing in the wheel assembly have caused failure of the bearing. Failure of the bearing has caused overload of the axle resulting in breakage of axle at the thread region.

11.3.5 Conclusion

The bearing has failed first due to frictional overheating, as a result of dry running conditions, causing bearing seizure. The failure of the bearing has caused overload of the axle and its breakage.



11.4 CASE STUDY: FAILURE OF COUPLING SHAFT AND BEVEL GEAR OF AN AEROENGINE

11.4.1 Background

One set of coupling shaft and bevel gear was rejected prematurely due to excess wear from aeroengine during overhaul micrometry checks. Maximum working temperature of the assembly was stated to be 150°C.

The coupling shaft was stated to have been made from low alloy steel of grade 38X2MUA as per GOST 2590-71 (TY-14-1-950-74). The shaft was hardened and tempered. The spline ends of the shafts were stated to have been nitrided to a depth of 0.15–0.30 mm. The ends of the shaft including the nitrided surface were plated with copper over a length of 30 ± 5 mm to a thickness of 0.003–0.006 mm. The surface of the shaft was phosphated. The specified chemical composition and mechanical properties of the coupling shaft are given in [Tables 11.4.1 and 11.4.2](#), respectively.

The bevel gear is stated to have been made using a low alloy steel of grade $12 \times 12\text{HB } \emptyset \text{ A-BD}$ as per GOST 2590-71 (TY 14-1-1885-76). The teeth of the gear and surface were carburized to a depth of 0.9–1.2 mm and hardened while the rest of the core region was only hardened and tempered. The drawing of the assembly did not specify any surface treatment to the splines cut within the gear which accommodates the coupling shaft. The specified chemical composition and mechanical properties of the bevel gear are given in [Tables 11.4.1 and 11.4.2](#), respectively.

Table 11.4.1 Chemical Composition in Wt.% of Coupling Shaft and Bevel Gear

	C	Si	Mn	Ni	Cr	Mo	V	Al	W	Fe
Coupling shaft	0.35–0.42	0.20–0.45	0.30–0.60	–	1.35–1.65	0.15–0.25	–	0.70–1.10	–	Rem.
Bevel gear	0.09–0.16	0.17–0.37	0.30–0.70	0.80–1.20	1.90–2.40	–	0.18–0.28	–	1.00–1.40	Rem.

Table 11.4.2 Mechanical Properties of Coupling Shaft and Bevel Gear

Properties	Coupling Shaft	Bevel Gear
	Heat treatment condition: Hardened (940°C) and Tempered (600–670°C)	Heat Treatment Condition: Hardened (910°C) and Tempered (550–560°C)
0.2% P.S (kg/mm ²)	80–85	90
UTS (kg/mm ²)	95–100	100
El (%)	15	12
CVN Energy (kgm/cm ²)	9–10	8
<i>Hardness</i>		
Case	Nitrided > 88 R _N	–
Core	285–341 HB	285–388 HB

11.4.2 Experimental Results

Visual Examination: The worn out bevel gear and coupling shaft in the as-received condition are shown in Figure 11.4.1. One of the splined ends of the shaft and the spline cut within the gear was observed to have been badly worn out (Fig. 11.4.2). The other end of the shaft was found to be intact.

Metallographic Examination: Longitudinal sections of both the shaft and the gear from the spline region were prepared for metallographic examination after edge protection by nickel plating. On etching with 2% nital solution the shaft sample showed a white etching layer of about 30 µm just below



Fig. 11.4.1 The worn out coupling shaft and bevel gear in the as-received condition.

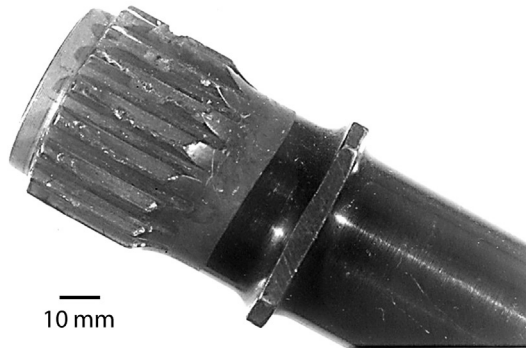


Fig. 11.4.2 The spline end of the coupling shaft showing the extent of wear and spalling.

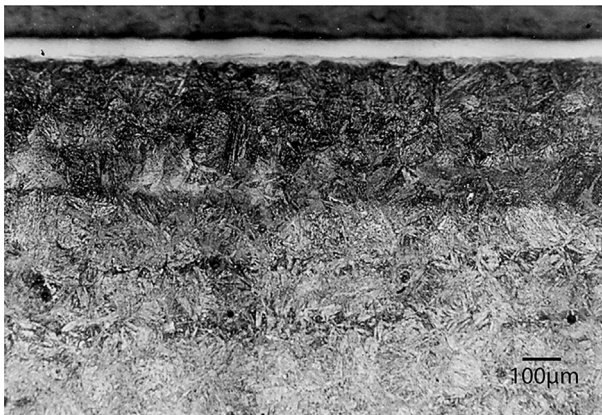


Fig. 11.4.3 Micrograph of longitudinal section of the coupling shaft showing a nitride case at the spline region. A white etching layer of about 20–30 μm at the surface is observed just below the nickel plating. Etchant: 2% nital.

the nickel plating (Fig. 11.4.3). A nitride case of about 300 μm depth was observed below the white etching layer as shown in Figure 11.4.3. The case showed a nitrided martensitic microstructure with prominent acicularity and a white film along the grain boundaries. The core of the sample showed a tempered martensite microstructure (Fig. 11.4.4).

The bevel gear sample showed a carburized case on the outer surface to a depth of about 350 μm . The spline cut within the gear was not found to have been given any surface treatment and it showed a tempered martensite microstructure which was same as the one seen on the entire section up to the outer surface (Fig. 11.4.5 and Fig. 11.4.6).



Fig. 11.4.4 Micrograph of longitudinal section of the coupling shaft showing a tempered martensite microstructure in the core. Etchant: 2% nital.

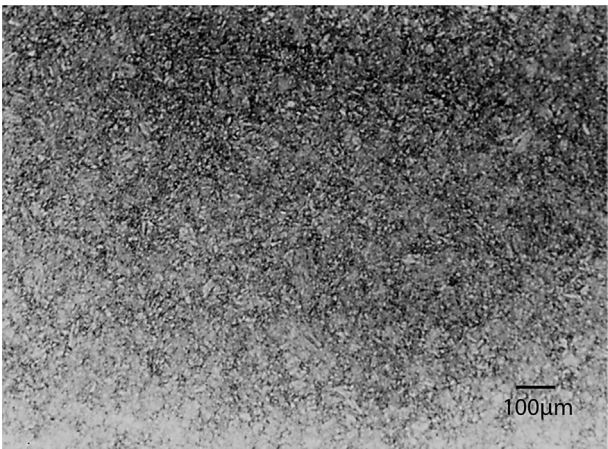


Fig. 11.4.5 Micrograph of longitudinal section of the bevel gear showing a tempered martensite microstructure in the case of the splined region. Etchant: 2% nital.

Hardness: Average Vickers hardness on different regions measured at 1 kg on surface and 30 kg in the core was as given here.

Sample	Location	Hardness HV
Shaft	Case	940
Shaft	Spline end core	315
Gear spline region	Surface	412
	Subsurface	408

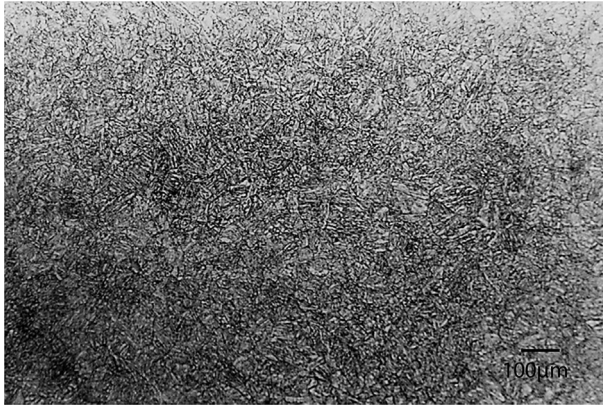


Fig. 11.4.6 Micrograph of longitudinal section of the bevel gear showing a tempered martensite microstructure in the core of the splined region. Etchant: 2% nital.

Microhardness scan of the shaft sample carried out at 100 g load was as given here.

Distance from surface (mm)	HV
0.03	1132
0.10	953
0.17	926
0.25	772
0.31	627
0.39	503
0.46	390
0.53	328
0.59	325

11.4.3 Chemical Analysis (Wt.%)

	C	Si	Mn	Cr	Mo	Al	Fe	Ni	W	V
Shaft	0.40	0.28	0.54	1.72	0.21	1.00	Bal.	—	—	—
Bevel Gear	0.14	0.30	0.46	1.90	0.05	—	Bal.	1.00	1.00	0.24

11.4.4 Discussion

Chemical analysis of the shaft sample suggested that it was made using the specified steel. The microstructure and hardness revealed that the spline end was nitrided which acquired the specified hardness of a nitrided case (940HV). The core of the shaft was found to have been heat treated as per the requirement of the specification.

Chemical analysis of the bevel gear sample suggested that it was made using the specified steel. Microstructure and hardness of the bevel gear indicated that it was case carburized except for the spline region. The spline region was simply hardened and tempered to a hardness level of 408HV which was much less as compared to the hardness of the coupling shaft (940HV) which gets engaged into it in the operation.

The metallurgical examination clearly suggested that the hard spline of the shaft came in contact with a relatively softer spline inside the gear causing wearing of the latter during service. The damage to the splined end of the shaft is a consequential effect aided by the presence of the brittle nitrided layer.

11.4.5 Conclusion

The spline region of the bevel gear was worn out primarily due to the wear caused by the harder splines of the shaft.



Elevated Temperature Failures

A time-dependent strain occurring under a constant stress and at a constant elevated temperature is known as 'creep'. When the strain occurring during creep does not relieve the stress, then the component under creep loading ruptures is called 'stress rupture'. The two major types of elevated temperature failures are creep and stress rupture. Creep begins in metals and alloys at a temperature slightly above the recrystallization temperature of that metal or alloy.

Grain boundary voids, cavities, and cracks at grain boundary triple points are microstructural evidences of creep damage. The presence of such microstructural degradation on either side of a fracture confirms stress rupture. Stress rupture can be either intergranular or transgranular. Mixed modes can also be found.

Alloys used in the hot environment of a system like gas turbine engines normally suffer from hot corrosion and erosion. Hot corrosion of alloys is accelerated oxidation caused by the presence of sodium chloride and sodium sulfate. The sources of such corrodants are the fuel used in engines, flue gases in a power plant, and the sea salt in the air. During the process of hot corrosion sulfur bearing slag forms on the surface of the component, which causes fluxing of protective oxide film on the alloy, allowing accelerated oxidation.

In order to protect surfaces from hot corrosion, firstly the component must be made of an alloy with improved hot corrosion resistance. The alloy has to be further supplemented by a protective coating. Aluminide-type coatings have been successfully tried on gas turbine blades and vanes against hot corrosion. Ni-Al on nickel base alloys, Co-Al on cobalt base alloys, Co-Cr-Al-Y and Fe-Cr-Al-Y alloy coatings are found beneficial to provide partial solution to the problem of hot corrosion.

Another type of elevated temperature failure is due to thermal fatigue, which is a deterioration of metallic material caused from temperature variations of the operating system. The degree of damage due to thermal fatigue depends on the magnitude and frequency of temperature fluctuation. Thermal fatigue cracks initiate at the surface and propagate inward. The cracking

due to thermal fatigue reveals a planar continuous surface. Thermal fatigue cracks can manifest itself in both intergranular and transgranular mode. Thermal fatigue can be minimized by taking suitable preventive methods in service; reducing large fluctuations in temperature of operating system, avoiding presence of stress concentrators in the design of the part, using suitable liners between parts having large temperature gradient, etc.



12.1 CASE STUDY: FAILURE OF A BOLT OF NOZZLE GUIDE VANE OF AN AEROENGINE

12.1.1 Background

It was reported that one bolt broke and fell down during testing of the aeroengine at engine test house. The subject bolt was meant to retain first stage nozzle guide vane between the casings along with stay rod. The bolt had completed 1007: 29 hours of service and its maximum working temperature was 1000°C.

12.1.2 Experimental Results

Visual Examination: The failed bolt in the as-received condition is shown in Fig. 12.1.1. The failure had taken place from the first thread root in the shank. The fracture surface was found to be coarse and crystalline.

Fractographic Examination: The fracture surface of the bolt was cleaned ultrasonically in an oxide remover followed by acetone and examined under scanning electron microscope (SEM). The fracture surface showed a granular feature at low magnification (Fig. 12.1.2). The fracture was essentially in intergranular mode showing secondary intergranular cracks when examined at higher magnification (Fig. 12.1.3).

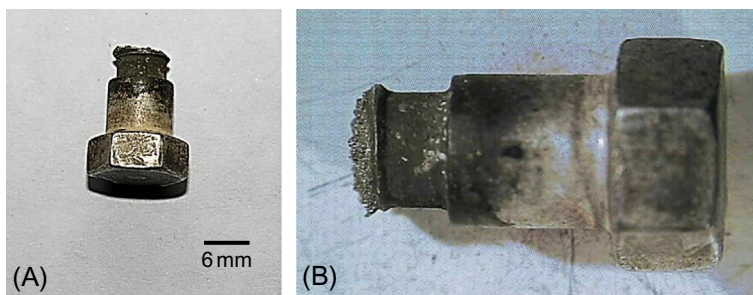


Fig. 12.1.1 Photograph showing the failed bolt in the as-received condition. (A) General view and (B) close-up showing coarse and crystalline fracture.

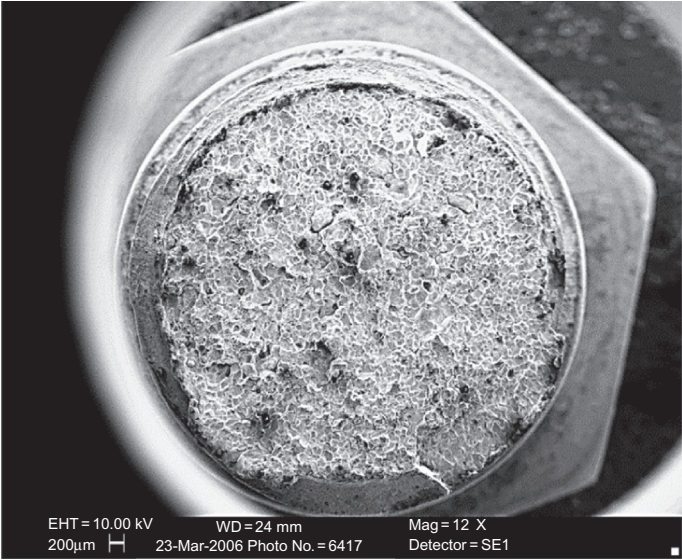


Fig. 12.1.2 SEM fractograph of the fracture surface showing granular feature at low magnification.

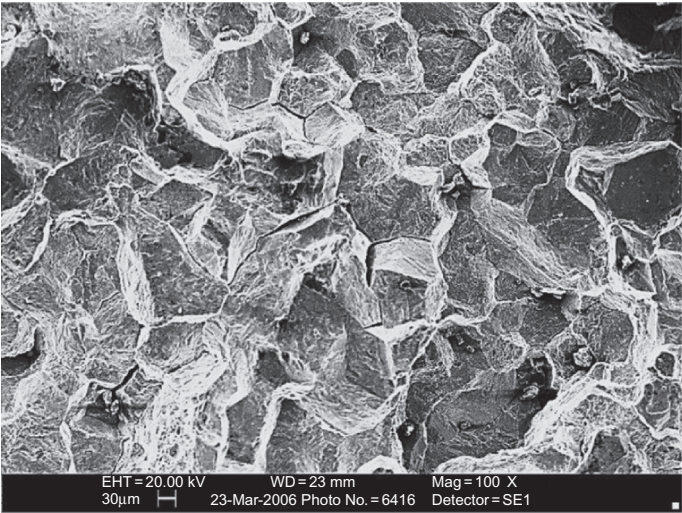


Fig. 12.1.3 SEM fractograph of the fracture surface showing essentially intergranular fracture with secondary intergranular cracks at higher magnification.

Metallographic Examination: Longitudinal section of the bolt containing the fracture surface was mounted and prepared for metallography. In the unetched condition, the sample showed a number of oxide and titanium carbo-nitride inclusions. On etching with Kalling's reagent the sample revealed primary cracking along the grain boundaries with secondary intergranular cracks (Fig. 12.1.4). The sample revealed equiaxed grains of gamma solid solution with annealing twins and carbides ($M_{23}C_6$) along the grain boundaries (Fig. 12.1.5). Grain coarsening was also observed

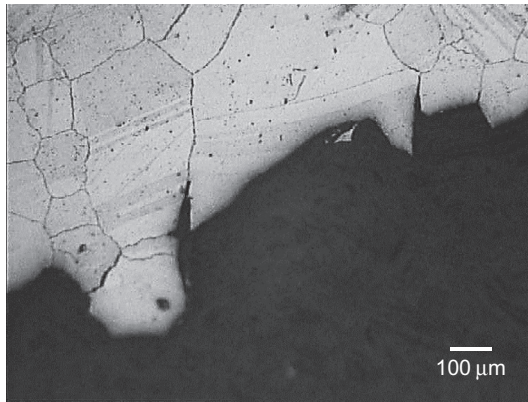


Fig. 12.1.4 Micrograph revealing primary cracking along the grain boundaries with secondary intergranular cracks. Etchant: Kalling's reagent

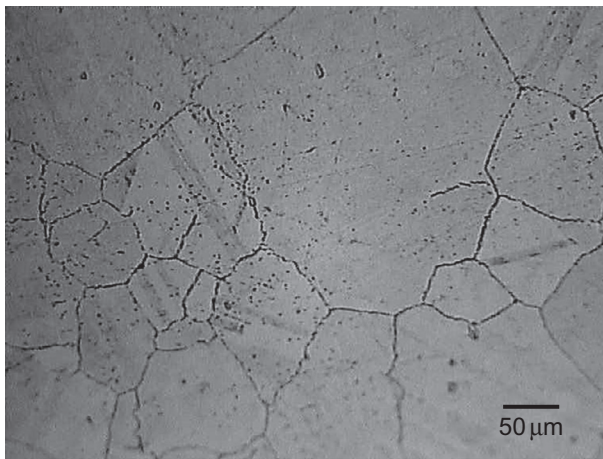


Fig. 12.1.5 Micrograph showing equiaxed and nonuniform grains of gamma solid solution with annealing twins and carbides ($M_{23}C_6$) along the grain boundaries. Etchant: Kalling's reagent.

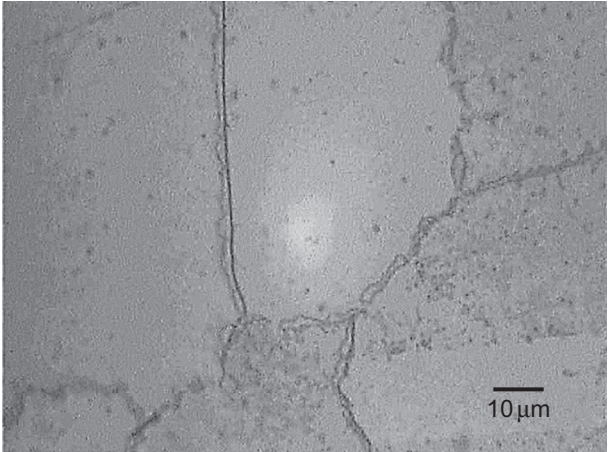


Fig. 12.1.6 Micrograph revealing continuous films of carbide ($M_{23}C_6$) along the grain boundaries. Etchant: Kalling's reagent.

following the process of Ostwald ripening. At higher magnification, a large number of grains revealed continuous films of carbide ($M_{23}C_6$) along the grain boundaries (Fig. 12.1.6). Gamma prime (γ') was not visible under optical microscope.

Hardness: Average Vickers hardness of the bolt material measured at 30 kg load was 320 HV.

12.1.3 Chemical Analysis (Wt.%)

Al	Si	Ti	Fe	Mo	Cr	Ni
1.44	1.30	2.69	0.45	0.84	21.17	Balance

12.1.4 Discussion

The chemical composition suggested that the bolt was a nickel base alloy nearly conforming to grade khN77TYuR of GOST 5632-72 and equivalent to Nimonic 80A. The microstructure and hardness indicated that the bolt was used in a solution-treated and age-hardened condition.

The microstructure revealed the presence of grain boundary films of carbides which must have formed either during heat treatment or during service at elevated temperature. The grain boundary film of carbides is detrimental to stress rupture life of nickel base alloys. It is known that the

bolt was used at elevated temperature. The rupture property of Nimonic 80A is as given here [1].

Temp °C	Stress Rupture (MPa)	
	100 h	1000 h
540	—	825
815	185	115
870	105	—

This data indicates that at temperatures above 800°C the rupture life is only 100 h at a stress as low as 185 MPa. Service at elevated temperature, presence of grain boundary carbide films, and intergranular mode of fracture point to the fact that the bolt had suffered stress rupture.

12.1.5 Conclusion

The bolt of nozzle guide vane had failed by stress rupture due to its application at elevated temperature for a service life of 1007 h.

12.1.6 Reference

- [1] Metals Handbook, vol. 3, Ninth Edition, ASM, Metals Park, Ohio, 1980, p. 222.



12.2 CASE STUDY: FAILURE OF VANES OF FRONT NOZZLE DIAPHRAGM ASSEMBLY OF AN AEROENGINE

12.2.1 Background

An aeroengine was withdrawn prematurely due to the notice of damage of vanes of front nozzle diaphragm assembly (FNDA). The FNDA, which was fitted between combustion chamber and compressor turbine assembly, was subjected to continuous flow of hot gases from combustion chamber at approximately 925°C. The aeroengine had completed 1047.53 h since last overhaul.

12.2.2 Experimental Results

Visual Examination: The damaged vanes in the as-received condition are shown in [Fig. 12.2.1](#). The breakage was observed on one-half to three-fourths of the distance from the upper end on the trailing edge of the vanes.



Fig. 12.2.1 Damaged vanes showing degradation and roughness on the surface of the convex side.



Fig. 12.2.2 Damaged vanes showing cracks and oxidation on the surface of the concave side.

The fracture surface was seen to be coarse and crystalline. The convex side of the vanes showed surface degradation similar to corrosion damage, surface roughness, and cracks (Fig. 12.2.1). Brown scale and some cracks also appeared on the concave side of the vanes (Fig. 12.2.2).

Fractographic Examination: Fracture surfaces of the vane were cleaned using oxide removing agent and acetone ultrasonically and observed under scanning electron microscope (SEM). Although the fracture surface of the

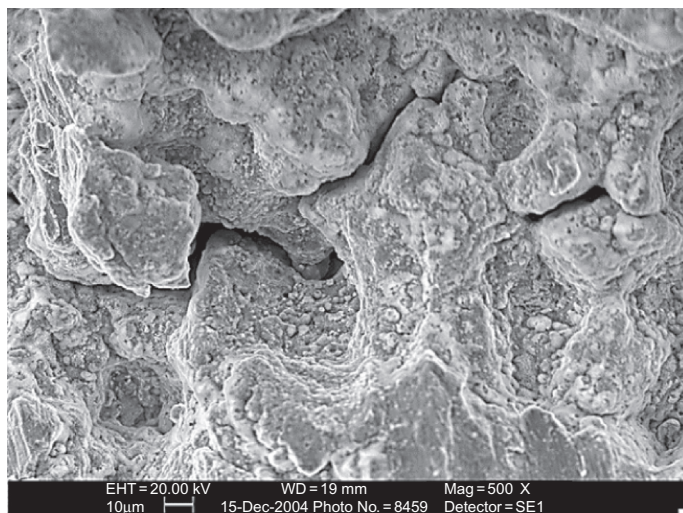


Fig. 12.2.3 SEM fractograph showing intergranular fracture with secondary cracks and oxide scale on the fracture surface.

vane sample was badly masked by oxide scale, some areas revealed intergranular fracture features with secondary intergranular cracks (Fig. 12.2.3).

Metallographic Examination: Samples were sectioned perpendicular to the long axis of a vane from three regions viz. the top aerofoil, the central aerofoil containing the fracture surface, and the root. The samples were prepared for metallography. The convex surface of the aerofoil region showed a surface coating of about 60 μm and a number of cracks in the coating propagating to the substrate (Fig. 12.2.4). The vertical cracks were observed to be joined by horizontal cracks and at places where the coating was found to be getting peeled off from the cracked portion (Fig. 12.2.5). At a number of locations on the concave side oxidation of the surface was seen (Fig. 12.2.6). On etching with Kalling's reagent the sample showed a cast (dendritic) microstructure (Fig. 12.2.7). The aerofoil region near the center of the vane showed zig-zag fracture surface and intergranular secondary crack in a cast structure (Fig. 12.2.8). The sample from near the root region showed the surface coating to be intact (Fig. 12.2.9). On etching, the sample revealed a cast microstructure with dendrites (Fig. 12.2.10).

The samples were observed under scanning electron microscope. Both the aerofoil and the root region showed predominately continuous films and at places discrete carbides along the grain boundaries (Figs. 12.2.11 and 12.2.12). At higher magnification the aerofoil region showed cuboidal



Fig. 12.2.4 Micrograph of the vane sample showing cracks on the surface coating on the convex side near the trailing edge of the aerofoil. Unetched.

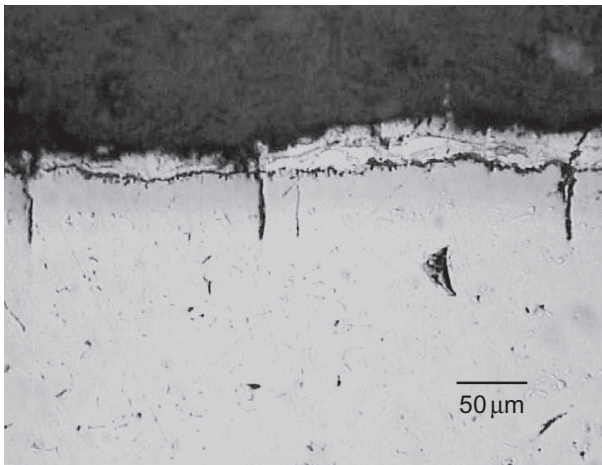


Fig. 12.2.5 Micrograph of the vane sample showing the process of peeling off of the surface coating on the convex side of the aerofoil. Unetched.

gamma prime in a gamma matrix (Fig. 12.2.13). The root region also revealed a similar microstructure but the gamma prime was considerably finer (Fig. 12.2.14).

Hardness: Average Vickers hardness values measured at 30 kg load was as given here.

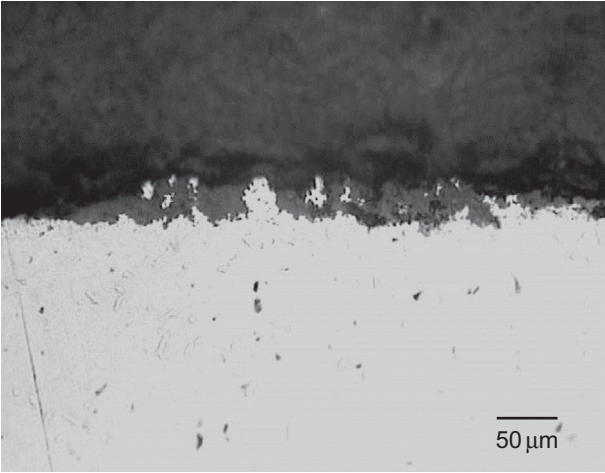


Fig. 12.2.6 Micrograph of the vane sample showing surface oxide scale on the concave side of the aerofoil. Unetched.

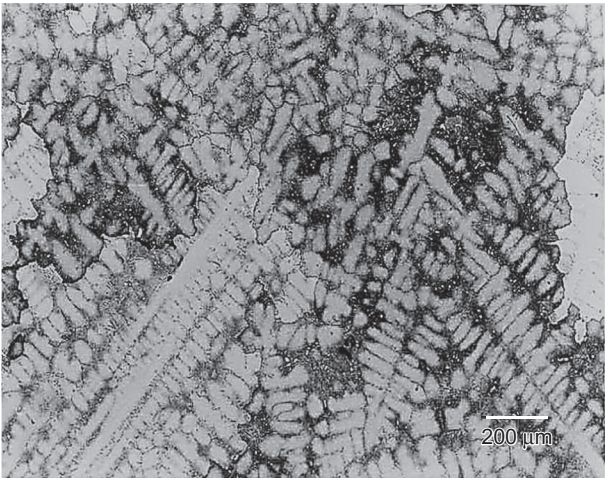


Fig. 12.2.7 Micrograph of the vane sample showing a cast (dendritic) microstructure in the aerofoil. Etchant: Kalling's reagent.

Location	Hardness (HV)
Aerofoil	342
Root	405

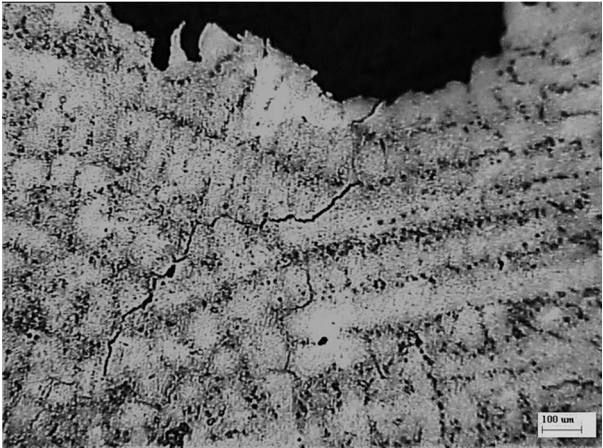


Fig. 12.2.8 Micrograph of the vane sample showing a zig-zag fracture surface and intergranular secondary crack in a cast microstructure. Etchant: Kalling's reagent.

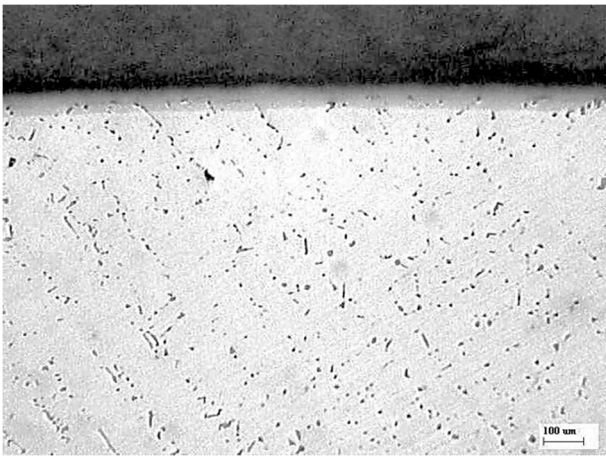


Fig. 12.2.9 Micrograph of the vane sample showing surface coating remaining intact near the root region. Unetched.

12.2.3 Chemical Analysis (Wt.%)

Al	Ti	Cr	Co	Ni
8.03	4.10	10.19	4.35	Base

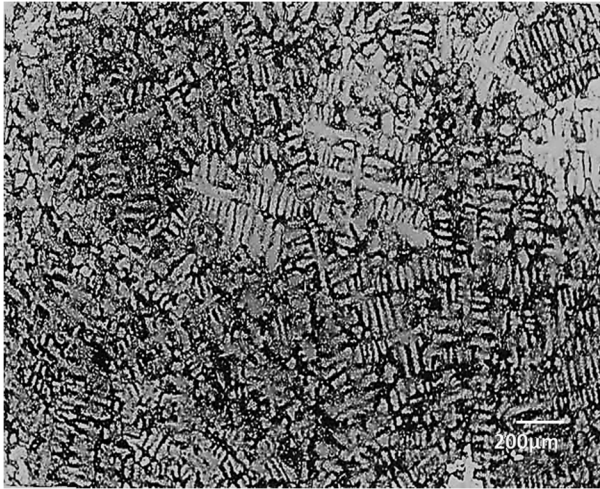


Fig. 12.2.10 Micrograph of the vane sample showing a cast dendritic microstructure in the root region. Etchant: Kalling's reagent.

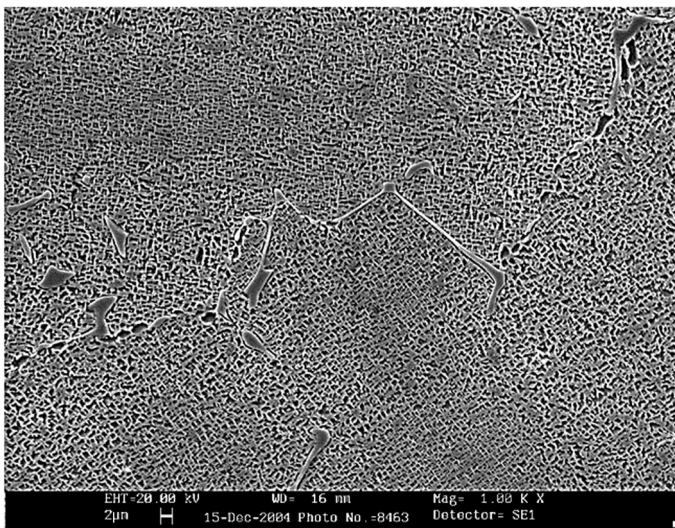


Fig. 12.2.11 SEM micrograph of the vane sample showing predominantly continuous films of carbide along the grain boundary and gamma prime within the grains of gamma in the aerofoil. Etchant: Kalling's reagent.

12.2.4 Discussion

Chemical analysis, microstructure, and hardness suggested that the vane was made of a nickel base alloy that did not conform to any standard and used in as-cast condition. The variation in the size of the gamma prime between the

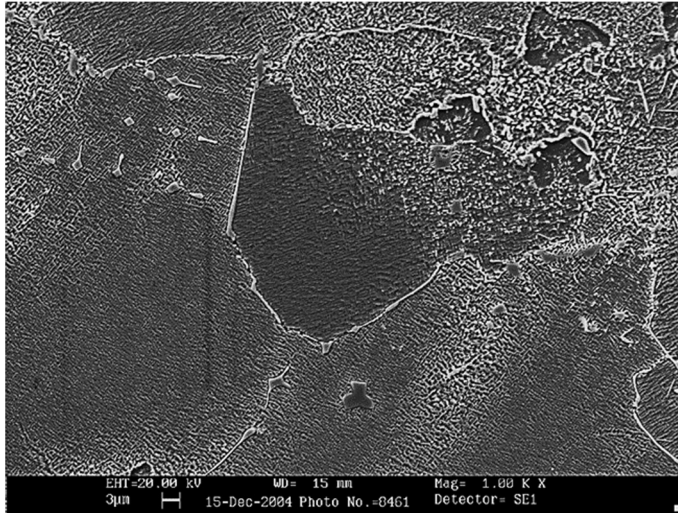


Fig. 12.2.12 SEM micrograph of the vane sample showing similar microstructure as in Fig. 12.2.11 in the root region. Etchant: Kalling's reagent.

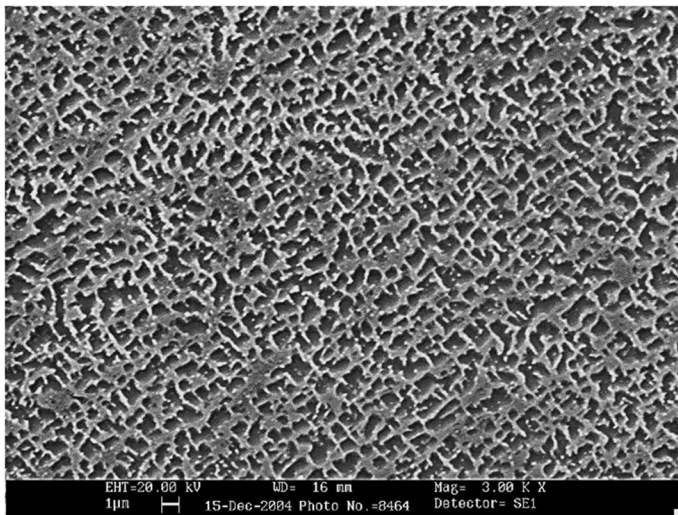


Fig. 12.2.13 SEM micrograph of the vane sample showing cuboidal gamma prime (γ') precipitates in a gamma (γ) matrix in the aerofoil. Etchant: Kalling's reagent.

aerofoil and root regions could most probably be due to (a) the difference in section thickness and consequently differential cooling rates after casting, and (b) overheating of the aerofoil region. The higher hardness observed in the root region as compared to the aerofoil was also due to the finer grain

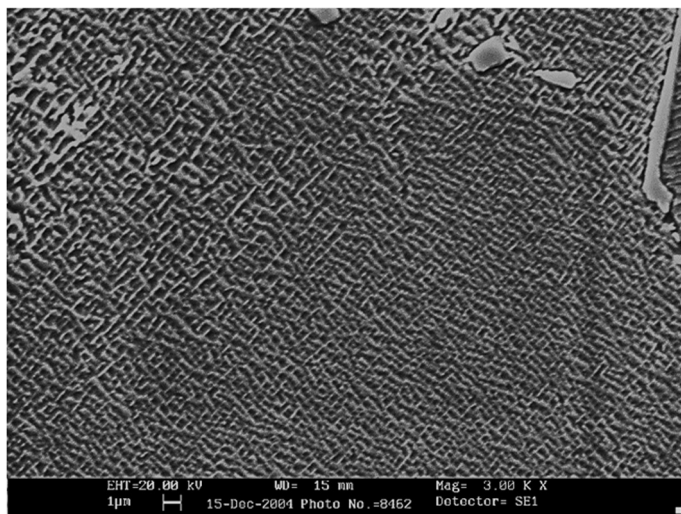


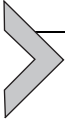
Fig. 12.2.14 SEM micrograph of the vane sample showing cuboidal gamma prime (γ') precipitates in a gamma (γ) matrix in the root region. Observe the finer size γ' precipitates as compared to [Fig. 12.2.13](#). Etchant: Kalling's reagent.

size of the root region. Continuous films of carbides along grain boundaries are not desirable in nickel base high-temperature alloys as this would affect hot ductility and reduce stress rupture life. The coating on the vane is expected to withstand the severe service conditions both thermal as well as mechanical up to an intended life. The coating on the convex side of the vane and especially near the trailing edge is observed to have cracked and peeled off at a number of locations. When the coating is strained at high temperature beyond its ductility limits it cracks. This allows penetration of hot combustion gases causing hot corrosion, oxidation, or both to the base metal and eventually causes spalling of the coating. The presence of intergranular cracking in the vanes confirmed the final failure to be due to stress rupture.

Impingement of hot blow of gases from combustion chamber at about 925°C which is probably focused on the trailing edge and convex side of the vanes has caused localized cracking and partial spalling of the surface coating resulting in overheating, hot corrosion, and oxidation of the basic material. This damage process has ultimately caused stress rupture of the vanes removing a considerable portion of the trailing edge.

12.2.5 Conclusion

The vanes had failed due to localized cracking and partial spalling of the surface coating resulting in overheating, hot corrosion, and oxidation of the base material ultimately causing stress rupture. Continuous films of carbide along the grain boundary have aided in the crack propagation and fracture of the vanes.



12.3 CASE STUDY: FAILURE OF CENTER MAIN BEARING OF AN AIRCRAFT ENGINE

12.3.1 Background

The center main bearing of an aircraft engine failed in service. The defect was noticed when there was an unusual noise from the engine during dry run. On checking it was found that there was no freedom of rotating assembly and no swing back. The defective bearing was made from steel to specification BSEM-832 by M/s Rolls Royce (UK). The maximum working temperature of the component was stated to be 120°C. The component had completed a service life of 2702.55 h, whereas its full life is not stated by the users. The center main bearing was used to support the rotating assembly.

12.3.2 Experimental Results

Visual Examination: The failed component in the as-received condition contained the bearing races, broken pieces of retainer, and severely worn out rollers (Fig. 12.3.1). The bearing race showed a large number of radial cracks of varying lengths. On opening the cracks, the fracture surfaces were found to be severely oxidized.

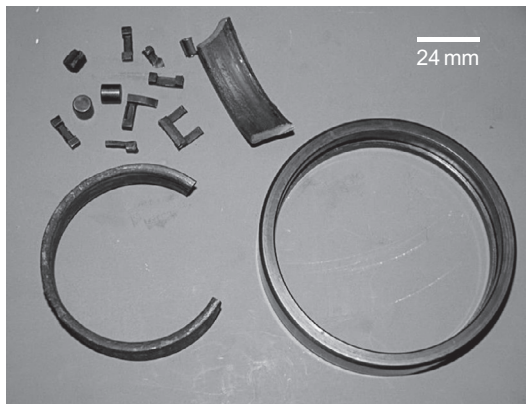


Fig. 12.3.1 Bearing components in the as-received condition.

Fractographic Examination: Fracture surface of a few freshly opened cracks from the bearing race was cleaned ultrasonically in acetone and examined under scanning electron microscope (SEM). The fracture surface showed a crystalline fracture having a number of radial secondary cracks (Fig. 12.3.2). The fracture surface revealed intergranular fracture with secondary cracks (Fig. 12.3.3). At higher magnification, the grain facets showed evidences of melting and resolidification (Fig. 12.3.4).

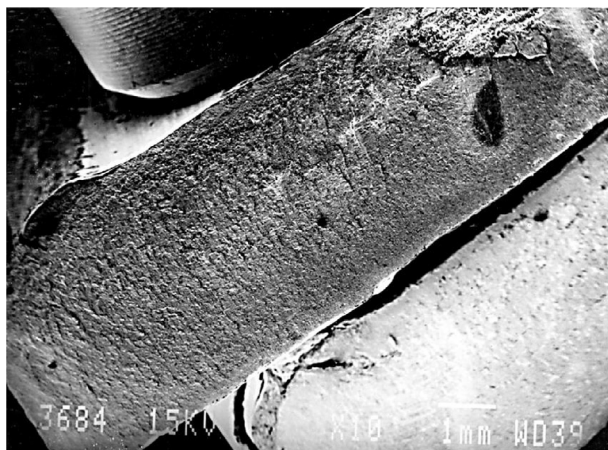


Fig. 12.3.2 SEM fractograph of the fracture surface of the bearing race showing a crystalline fracture having a number of radial secondary cracks.

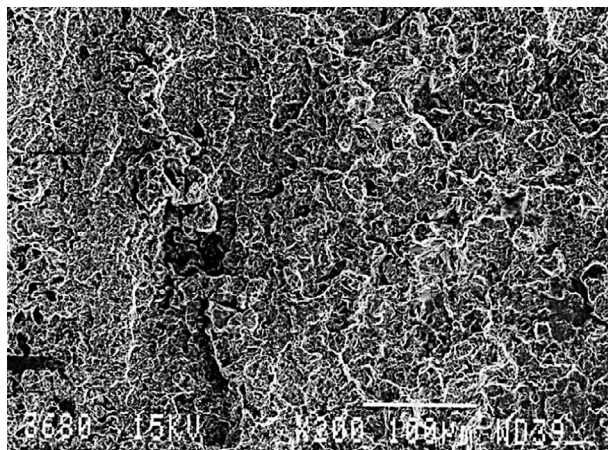


Fig. 12.3.3 SEM fractograph of the fracture surface of the bearing race showing intergranular fracture with secondary cracks.

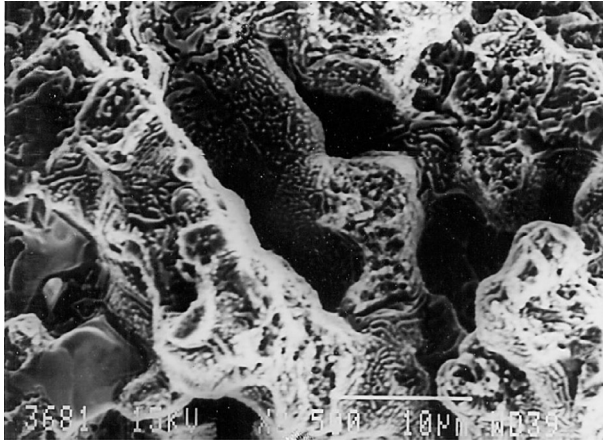


Fig. 12.3.4 SEM fractograph of the fracture surface of the bearing race showing dendrites.



Fig. 12.3.5 Micrograph of longitudinal section of the race sample showing a number of nonmetallic inclusions. Unetched.

Metallographic Examination: Longitudinal and transverse sections of the sample from the bearing race were prepared for metallographic examination. In the unetched condition, the race sample showed a large population of nonmetallic inclusions (Fig. 12.3.5). Race sample near the fracture surface showed voids and fissures connected by intergranular cracks (Fig. 12.3.6). Intergranular cracks were also found deep inside the sample (Fig. 12.3.7). Wedge-shaped micro-cavities were also noticed in many parts of the sample (Fig. 12.3.8). On etching with 2% nital, the race sample revealed evidences

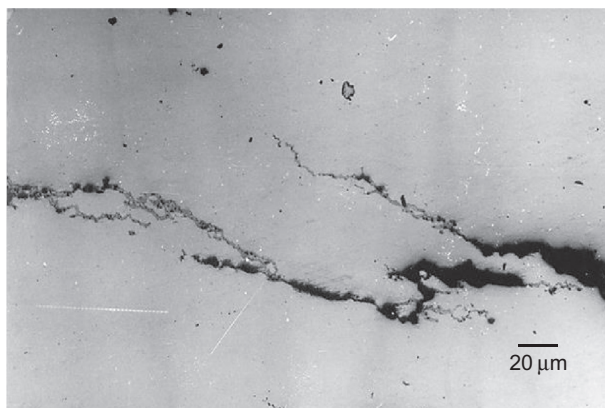


Fig. 12.3.6 Micrograph of longitudinal section of the race sample near the fracture region showing voids and fissures connected by cracks. Unetched.

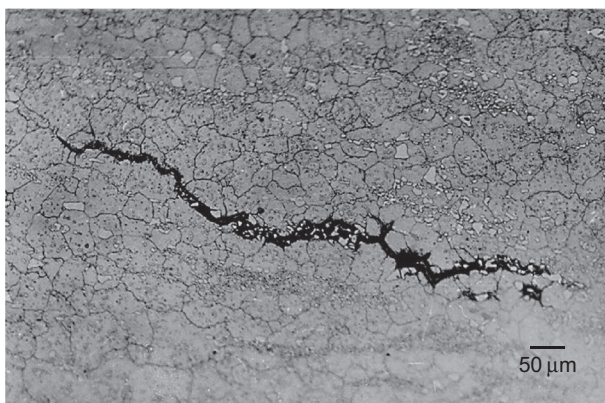


Fig. 12.3.7 Micrograph of a section of the race sample away from the fracture region showing intergranular cracks inside the material. Etchant: 2% nital.



Fig. 12.3.8 Micrograph of a section of the race sample showing wedge-shaped microcavities. Unetched.

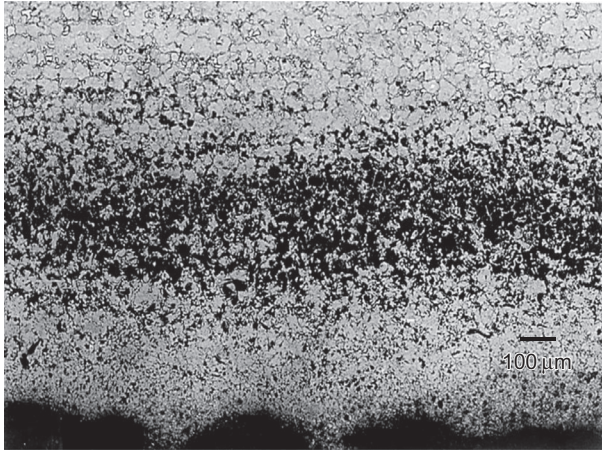


Fig. 12.3.9 Micrograph of a section of the race sample near the inner surface showing pitting and oxidation. Etchant: 2% nital.

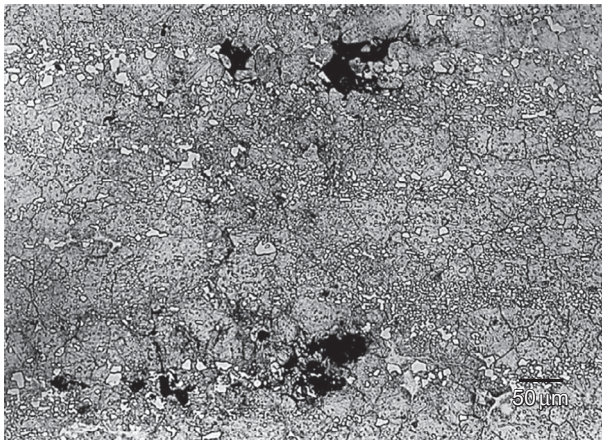


Fig. 12.3.10 Micrograph of longitudinal section of the race sample showing intergranular cavities and undissolved carbides in a matrix of tempered martensite. Etchant: 2% nital.

of pitting and oxidation near the inner surface (Fig. 12.3.9). Longitudinal section of the race sample showed intergranular cavities and undissolved carbides in a matrix of tempered martensite (Fig. 12.3.10). Transverse section revealed carbide precipitation along the grain boundaries (Fig. 12.3.11).

Hardness: Microhardness scan at 200 g load on the race sample from the inner surface is given here.

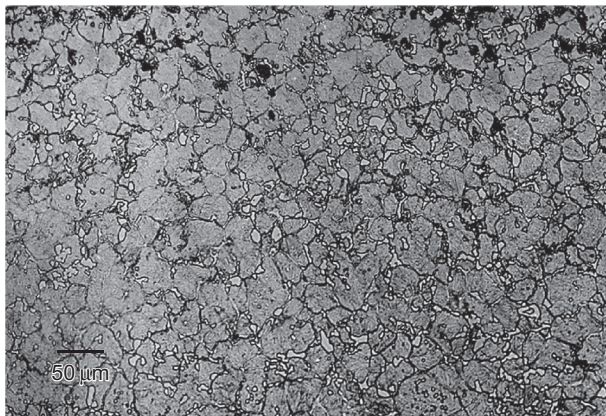


Fig. 12.3.11 Micrograph of transverse section of the race sample showing carbide precipitation along the grain boundaries. Etchant: 2% nital.

Near the inner surface: 830 HV

At the pitted and oxidized zone: 704 HV

At the carbide-segregated zone: 736 HV

Parent material: 803 HV

Average Vickers hardness of the bearing race measured at 10 kg load was 803 HV.

12.3.3 Chemical Analysis of Bearing Race (Wt.%)

C	Si	Mn	W	Cr	V	Mo	S	Fe
0.75	0.33	0.34	18.66	4.24	1.32	0.40	0.011	Rem.

12.3.4 Discussion

Chemical composition and microstructure suggested that the bearing race was made from W-Cr-V-Mo steel of the specified grade and used in hardened and tempered condition. Microstructure of the race sample indicated that it was heated to a very high temperature during service causing microstructural changes near the inner surface. Wedge-shaped cavities, voids, and intergranular cracking are also evidences of overheating conditions the bearing race was subjected to. The structural changes and variations in hardness from inner to the outer surface of the race sample could have resulted due to high frictional heat generated due to lack of lubrication. Rotational creep of the bearing probably has caused large number of radial

crackings in the race. The heat generated due to frictional heat has raised the race temperature to melting temperature of the steel at localized regions as evidenced by formation of dendrites as a result of melting and resolidification.

12.3.5 Conclusion

Bearing race of the center main bearing had failed in service as a result of frictional overheating due to lack of lubrication.



12.4 CASE STUDY: FAILURE OF JET NOZZLE CASING OF AN AEROENGINE

12.4.1 Background

At one of the operational bases, a jet nozzle casing of an aeroengine got ruptured during flying at augmented rating (after burner). It was stated that the jet nozzle casing was made from a material grade 8-6, designation KhN78T of specification GOST 5632-72 (Russian Specification). The maximum working temperature of the component was 950°C. The component had completed 1336:16 hours while the total technical life was 1800:00 hours.

12.4.2 Experimental Results

Visual Examination: The jet nozzle casing in the as-received condition is shown in Fig. 12.4.1. The component was observed to have suffered a



Fig. 12.4.1 Failed jet nozzle casing in the as-received condition.

longitudinal rupture of size about 400 mm. The fracture surface was found to have been badly oxidized and the scale was not easily removable even by chemical cleaning.

It was difficult to locate the origin of failure. Hence a few samples were cut from the fracture surface of the failed jet nozzle casing and cleaned with oxide removing agent followed by acetone cleaning. The fracture surfaces were examined under scanning electron microscope (SEM). The fracture mode could not be observed as the fracture surfaces contained thick adherent oxide scale which was not removable by available cleaning agents.

Metallographic Examination: Sections of the samples near and away from the fracture were prepared for metallographic examination. In the unetched condition, the jet nozzle casing sample near the fracture showed severe grain boundary oxidation and cracks. On etching with aqua regia, the sample revealed intergranular oxidation and cracks near the surface and equiaxed grains (Fig. 12.4.2). Section away from fracture also revealed grain boundary oxidation and cracks near the surface and equiaxed grains (Fig. 12.4.3).

Hardness: Average Vickers hardness measured at 10kg load was as given here.

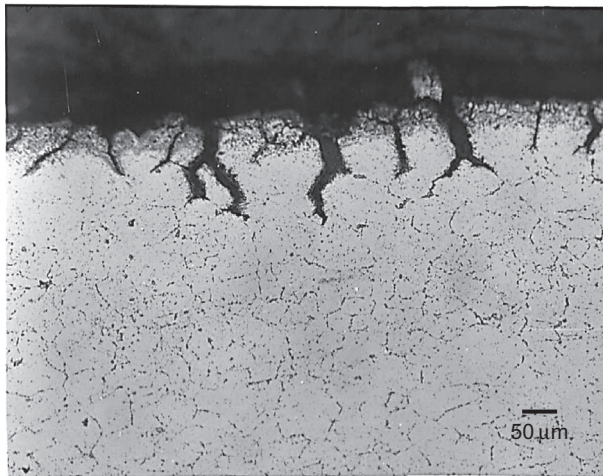


Fig. 12.4.2 Micrograph of a section of the sample from failed region showing grain boundary oxidation and cracks near the surface and equiaxed grains. Etchant: Aqua regia.

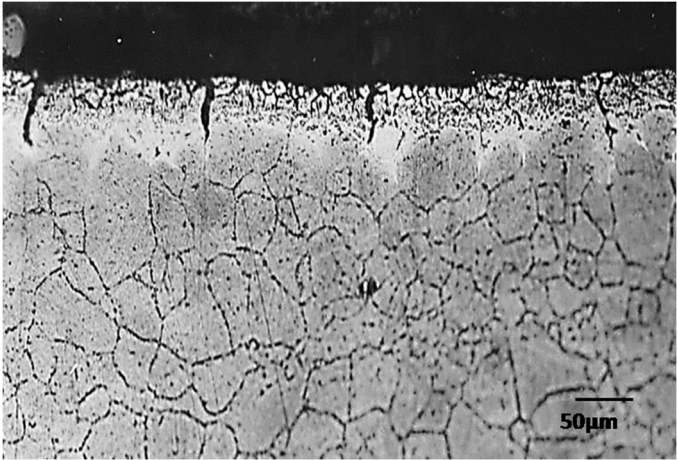


Fig. 12.4.3 Micrograph of a section of the sample away from failed region showing grain boundary oxidation and cracks near the surface and equiaxed grains. Etchant: Aqua regia.

Location	Hardness (HV)
Away from failure	325
Near failure	285

Evaluation of Hot Corrosion: The hot corrosion damage is evaluated from the micrograph as shown in [Fig. 12.4.4](#). [1]

12.4.3 Chemical Analysis (Wt%)

C	Cr	Si	Mn	Ti	Al	Ni
0.072	20.7	0.69	0.57	0.19	0.09	Base

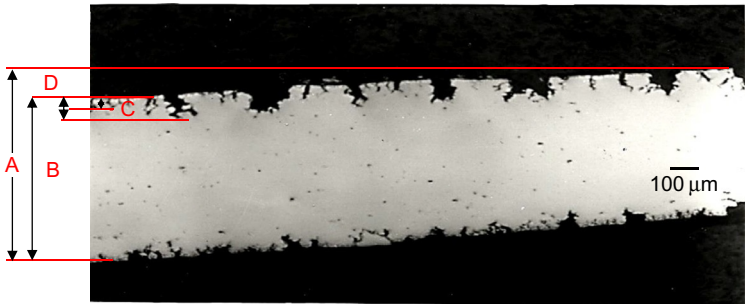


Fig. 12.4.4 Micrograph of a section showing hot corrosion damage near the surface. Unetched.

Metal loss per side $= (A - B)/2 = (800 - 700)/2 = 50 \text{ } \mu\text{m}$

Continuous penetration per side, $C = 70 \text{ } \mu\text{m}$

Maximum penetration per side, $D = 100 \text{ } \mu\text{m}$

Total metal affected per side $= (A - B)/2 + D = 50 + 100 = 150 \text{ } \mu\text{m}$

12.4.4 Discussion

The failed jet nozzle casing was made from Ni-Cr alloy of the specified grade. In the absence of any information on the intended microstructure and mechanical properties, it is difficult to comment on the same in the case of the failed component. Nevertheless, it is speculated that the alloy might have been used in annealed condition as evidenced by single-phase equiaxed grains in metallography. The alloy must have developed the hardness of about 300 HV (average) due to aging in service during which the presence of traces of titanium and aluminum might have formed γ -prime precipitates with nickel. Precipitation of chromium carbide could also be another reason for the development of hardness in the component.

Such alloys are generally recommended to be used up to 1000°C. Intense scaling due to high-temperature oxidation initiates between 1100°C and 1150°C in these alloys. Metallographic examination revealed the presence of severe intergranular oxidation and cracks near the surface. Since the alloy is recommended for high-temperature application, the surface damage could be due to hot corrosion rather than simple overheating. Evaluation of hot corrosion damage showed that the total metal affected per side was about 150 μm which is alarming considering the fact that the total thickness of the jet nozzle casing was only 800 μm . The samples both near and away from the failed region revealed nonuniform grain size suggesting overheating of the material.

The deterioration indicated that the casing was subjected to hot corrosion, resulting in considerable loss of material and failure of the component in intergranular mode. Lower hardness of the material near failure could be due to localized grain coarsening.

12.4.5 Conclusion

Cracking of the jet nozzle casing was a result of hot corrosion of the component.



12.5 CASE STUDY: FAILURE OF TURBINE STATOR VANES OF AN AEROENGINE

12.5.1 Background

Two turbine stator vanes revealed cracks on their surfaces during strip examination of an aeroengine. The total technical life of the vanes was decided on their condition. Two segments of stator vanes were received for investigation.

12.5.2 Experimental Results

Visual Examination: The turbine stator vanes in the as-received condition are shown in Fig. 12.5.1. Each segment contained six hollow vanes. Three vanes on one segment and two vanes on another segment showed cracks near its trailing edges (Figs. 12.5.2 and 12.5.3). The cracks were 6–10 mm long. The surfaces of the vanes were found to be rough. The cracked portions of the vanes were sectioned using an EDM wire cut machine and removed. On opening the cracks, the fracture surfaces were found to contain dark adherent oxide.

Fractographic Examination: The fracture surfaces were cleaned using an oxide removing solution, followed by cleaning in acetone ultrasonically. On examining under SEM, the fracture surfaces showed thick adherent oxide scale, which probably was not removed by cleaning (Figs. 12.5.4 and 12.5.5). A typical energy dispersive X-ray spectroscopy (EDAX) of the fracture surface is shown in Fig. 12.5.6.

Metallographic Examination: Longitudinal and transverse sections from the aerofoil and root regions of the vanes were prepared for metallographic examination.

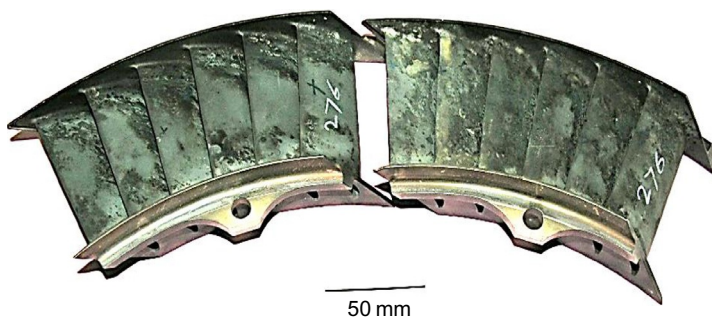


Fig. 12.5.1 Two segments of failed vanes in the as-received condition.



Fig. 12.5.2 Close-up view of one segment of the vanes showing cracks near the trailing edge on the convex side of the aerofoil.



Fig. 12.5.3 Close-up view of another segment of the vanes showing cracks near the trailing edge on the convex side.

In the unetched condition, the longitudinal section of the aerofoil of the vane samples showed oxide-filled cracks near the surface (Fig. 12.5.7A). On etching with Kalling's reagent, the sample revealed that the oxide-filled surface cracks predominantly propagated through interdendritic/grain boundary regions (Fig. 12.5.7B). Sections of the sample taken from both aerofoil and the root regions showed dendritic structure (Fig. 12.5.8). The structure revealed primary eutectic carbides (M_7C_3) along the interdendritic regions and mottled (fine spotted) chromium-rich ($M_{23}C_6$) carbides in a matrix of solid solution of cobalt on both aerofoil as well as root regions (Fig. 12.5.9).

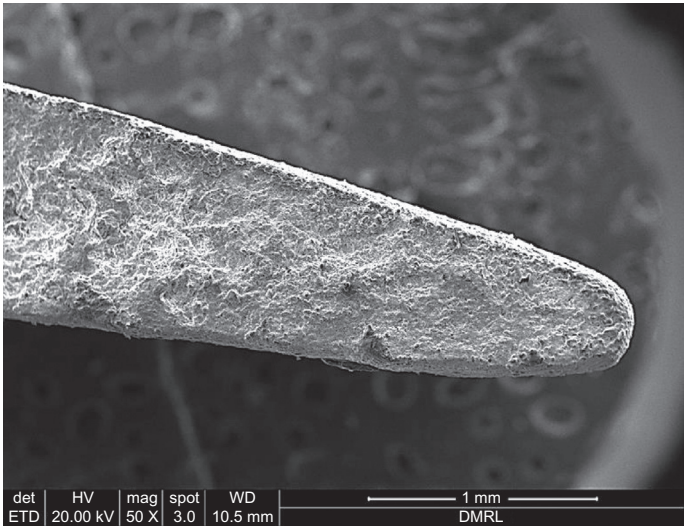


Fig. 12.5.4 SEM fractograph of one of the typical fracture surfaces at the trailing edge of the aerofoil.

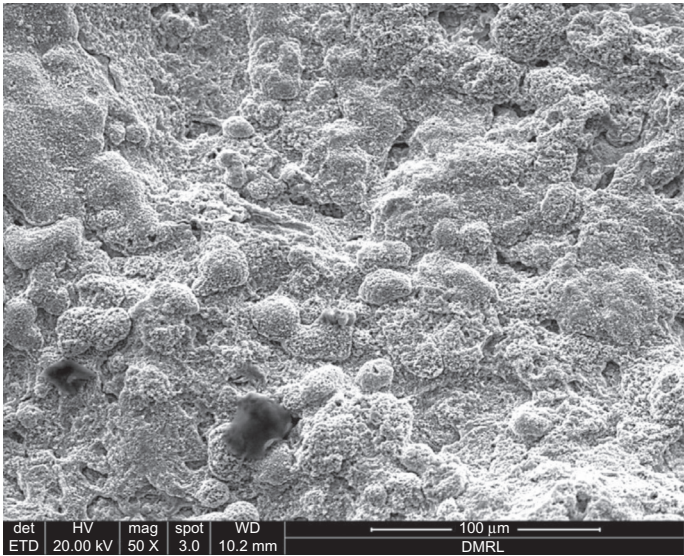


Fig. 12.5.5 SEM fractograph of the fracture surface showing thick adherent oxide scale.

Hardness: Average Vickers hardness of the sample measured at 30 kg load was as given here.

Aerofoil: 390 HV

Root: 335 HV

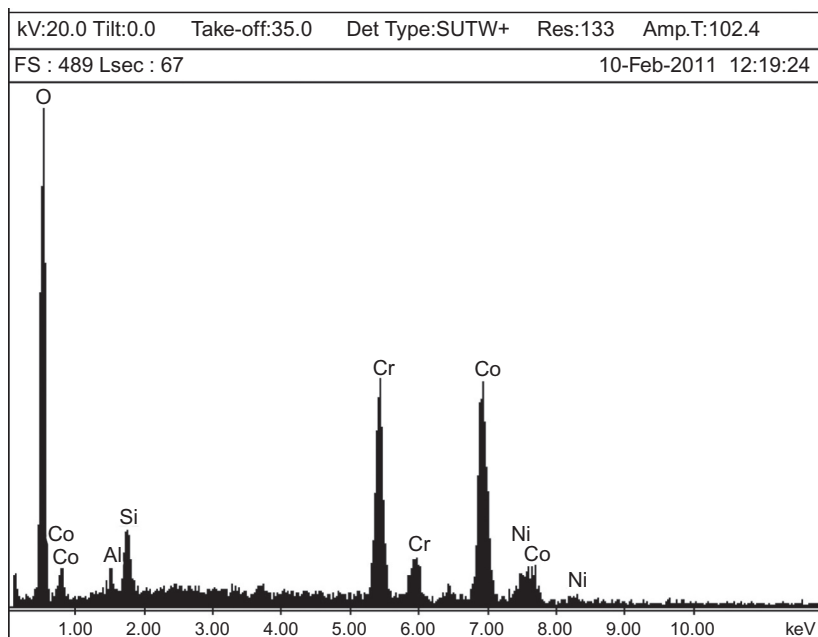


Fig. 12.5.6 EDAX spectrum of the oxide scale found on the fracture surface.

12.5.3 Chemical Analysis (Wt.%)

C	W	Cr	Ni	Co
0.50	11.25	27.20	11.64	Base

12.5.4 Discussion

Chemical composition and microstructure of the failed turbine stator vane suggested that it was a cobalt base alloy casting nearly conforming to grade FSX-414 and Mar-M509 (American specification). Hardness of the aerofoil and root indicated that the part was used in as-cast and aged condition. Metallographic examination of longitudinal section near the cracked area in the aerofoil revealed several oxide-filled cracks, which were characteristic thermal fatigue cracks.

In high-temperature cobalt base alloys, surface oxidation at elevated temperature can deplete alloying elements (notably chromium) in the base metal, thereby reducing strength of the metal and making it susceptible to thermal fatigue cracking [1]. The oxide prevents the crack from closing,

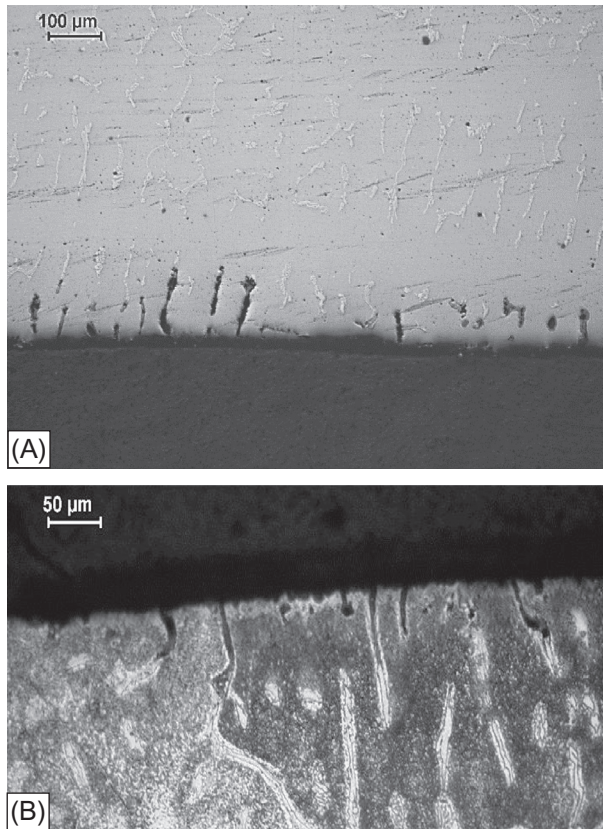


Fig. 12.5.7 Micrograph of the longitudinal section from the aerofoil of a vane sample showing oxide-filled grain boundary cracks near the surface. (A) Unetched and (B) etched: Kalling's reagent.

and additional stresses are imposed on the crack tip, which may accelerate intergranular cracking. Age hardening occurs in cobalt base alloys due to intragranular precipitation of carbide particles ($M_{23}C_6$) that were induced by engine operating temperatures. Age hardening temperature for obtaining useful mechanical properties for this alloy is 760°C after solution annealing at $1090\text{--}1220^{\circ}\text{C}$. This aging temperature promotes the formation of discrete $M_{23}C_6$ carbides in the matrix. Aging continues at higher temperature resulting in acicular or lamellar precipitate formation of $M_{23}C_6$ carbides, which makes the alloy extremely brittle [2]. In addition, precipitation of chromium carbides necessarily depletes the matrix of chromium an essential constituent for oxidation resistance.

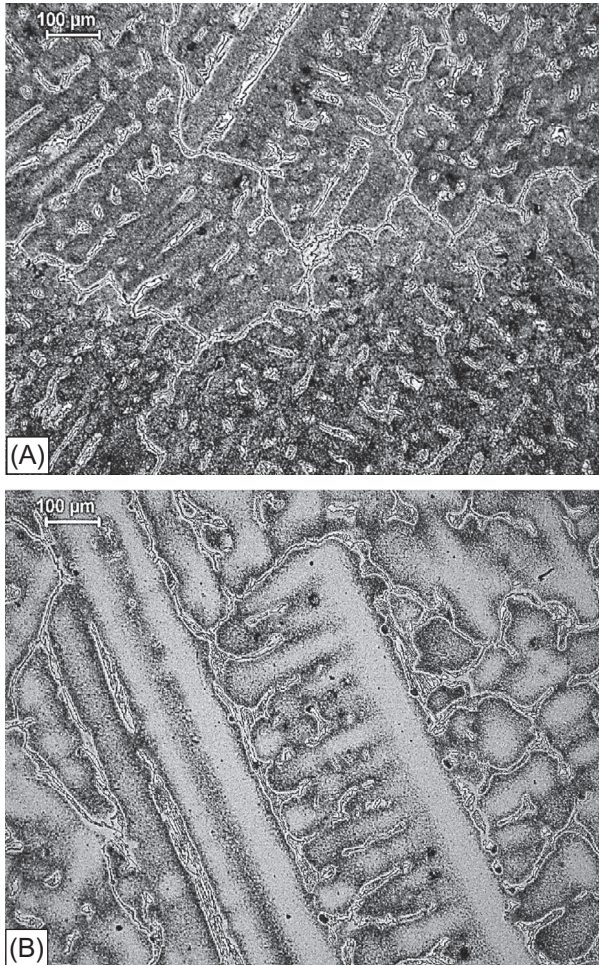


Fig. 12.5.8 Micrographs of the vane sample showing a dendritic structure. (A) Aerofoil and (B) root. Etchant: Kalling's reagent.

The roughness on the surface of the vanes indicated its prolonged use at high temperature. The surface did not reveal the presence of any thermal barrier coating. Thick adherent oxide was found on the cracked surfaces of the vane. Interdendritic oxide-filled surface cracks were found in large numbers. The higher hardness of the aerofoil as compared to the root region of the vane confirmed age hardening of the aerofoil region due to high operating temperature. The alloy being susceptible to continuous age hardening at higher

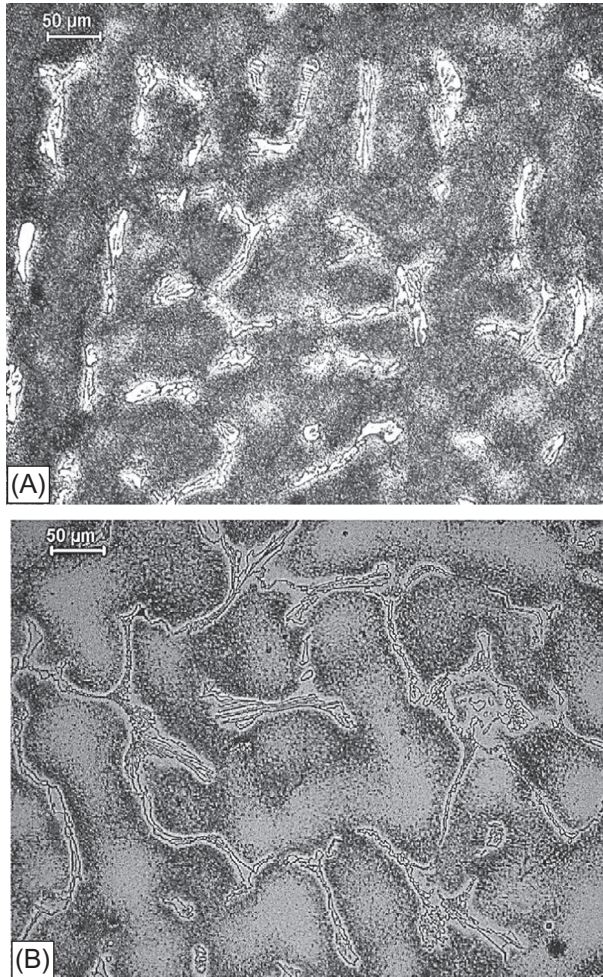


Fig. 12.5.9 Micrograph of a section of the vane sample showing carbide particles and eutectic along the grain boundaries and mottled (fine spots) chromium-rich carbides in a matrix of solid solution of cobalt. (A) Aerofoil and (B) root. Etchant: Kalling's reagent.

temperature ($>760^{\circ}\text{C}$) resulted in low ductility. All the evidences point to the fact that the vanes had suffered a thermal fatigue cracking in service.

12.5.5 Conclusion

Cracking of the aerofoil region of the vanes was caused by thermal fatigue. The thermal fatigue had resulted due to interdendritic/grain boundary

oxidation, low ductility due to age hardening of the alloy and thermal stress cycles the vanes were subjected to, during its service.

12.5.6 References

- [1] Metals Handbook, vol. 10, 8th ed., American Society for Metals, Metals Park, Ohio, p. 272.
- [2] Metals Handbook, vol. 2, 10th ed., American Society for Metals, Metals Park, Ohio, p. 1404.

BIBLIOGRAPHY

- [1] Practical Engineering Failure Analysis, Hani M. Tawancy, Anwar Ul-Hamid, Nureddin M. Abbas, Marcel Dekker, New York 10016, 2004.
- [2] D.J. Wulpi, Understanding How Components Fail, ASM International, Metals Park, OH, 2001.
- [3] S. Bhattacharya, V.E. Johnson, S. Agarwal, M.A.H. Howes (Eds.), IITRI Fracture Handbook: Failure Analysis of Metallic Materials by Scanning Electron Microscopy, Metals Research Division, Illinois Institute of Technology Research, Chicago, 1979.
- [4] B. Ross, Investigating Mechanical Failures, Chapman and Hall, London, UK, 1995.
- [5] P.M. Unterweiser (Ed.), Case Histories in Failure Analysis, ASM, Metals Park, OH, 1979.
- [6] D.R.H. Jones, Materials Failure Analysis, Pergamon Press, London, UK, 1993.
- [7] F.R. Hutchings, P.M. Unterweiser (Eds.), Failure Analysis: The British Engine Technical Reports, ASM, Metals Park, OH, 1981.
- [8] V.J. Colangelo, F.A. Heiser, Analysis of Metallurgical Failures, 2nd ed., John Wiley and Sons, Wiley-Interscience, New York, 1987.
- [9] A. Venugopal Reddy, Investigation of Aeronautical and Engineering Component Failures, CRC Press, New York, 2004.
- [10] W.T. Becker, R.J. Shipley (Eds.), ASM Metals Handbook vol. 11, Failure Analysis and Prevention, ASM International, Materials Park, OH 44073-0022, 2002.
- [11] K. Mills. ASM Metals Handbook vol. 12, 9th ed., Fractography, ASM International, Materials Park, OH 44073-0022, 1992.
- [12] B. Raj, T. Jaykumar, M. Thavasimuthu, Practical Non Destructive Testing, 2nd ed., Woodhead Publishing Limited, Cambridge, 2002.
- [13] McMaster, C. Robert, Non Destructive Testing Handbook, vol. II, Ronald Press, New York, 1963.
- [14] J.L. McCall, P.M. French (Eds.), Metallography in Failure Analysis, Plenum Corp., New York, 1978.
- [15] G.F. Vandervoort, Metallography as Quality Control Tool, Plenum Press, New York, 1979.
- [16] G.F. Vander Voort, (Ed.), ASM Metals Handbook vol. 9, Metallography and Microstructures, ASM International, Materials Park, OH 44073-0022, 2004.
- [17] G.E. Dieter, Mechanical Metallurgy, 3rd ed., Mc Graw-Hill, London, 1986.
- [18] N.E. Dowling, Mechanical Behaviour of Materials: Engineering Methods for Deformation, Fracture and Fatigue, 2nd ed., Prentice Hall, Upper Saddle River, NJ, 1999.
- [19] R. Steiner, ASM Metals Handbook vol. 1, 10th ed., Properties and Selection: Irons, Steels and High Performance Alloys, ASM International, Materials Park, OH 44073-0022, 1992.
- [20] ASM International Handbook Committee. ASM Metals Handbook vol. 2, 10th ed., Properties and Selection: Non-Ferrous Alloys and Special Purpose Materials, ASM International, Materials Park, OH 44073-0022, 1992.
- [21] R. Kiessling, Non-Metallic Inclusions in Steels, Part III, vol. 115, Iron and Steel Institute Publication, London, 1968.
- [22] Steels: Heat Treatment and Processing Principles, George Krauss, ASM International, Metals Park, OH 44073, 1990.
- [23] Heat Treater's Guide Standard Practices and Procedures for Steel, Paul M. Unterweiser, Howard E. Boyer and James J. Kubbs, ASM International, Metals Park, OH 44073, 1985.

- [24] V. Singh, Heat Treatment of Metals, 2nd ed., Standard Publishers Distributors, Delhi, 2004.
- [25] ASM International Handbook Committee. ASM Metals Handbook vol. 4, 10th ed., Heat Treating, ASM International, Materials Park, OH 44073-0022, 1991.
- [26] L.M. Gourd, Principles of Welding Technology, Edward Arnold, London, 1986.
- [27] R.A. Higgins, Engineering Metallurgy (Part II), Edward Arnold, London, 1986.
- [28] J.F. Lancaster, Metallurgy of Welding, Woodhead Publishing Limited, Cambridge, England, 1999.
- [29] C.R. Brooks, A. Choudhary, Metallurgical Failure Analysis, Mc Graw-Hill, New York, 1993.
- [30] R.D. Barer, B.F. Peters, Why Metals Fail, Gordon and Breach, New York, 1970.
- [31] W.D. Pilkey, R.E. Peterson, J.E. Peterson, K.M. Clark, Petersons Stress Concentration Factors, 2nd ed., John Wiley and Sons, New York, 1997.
- [32] J.A. Bannantine, J.J. Corner, J.L. Handrock, Fundamentals of Metal Fatigue Analysis, Prentice-Hall, Englewood Cliffs, NJ, 1993.
- [33] M.G. Fontana, N.D. Green, Corrosion Engineering, 2nd ed., Mc Graw-Hill, New York, 1978.
- [34] Material Selection for Corrosion Control, Chawla S.L and Gupta R.K., ASM International, Metals Park, OH 44073, 1993.
- [35] G.M. Ugiansky, J.H. Payer (Eds.), Stress Corrosion Cracking, ASTM International, West Conshohocken, PA, 1979 STP 665.
- [36] P. Cotterill, Hydrogen Embrittlement of Metals, Pergamon, New York, 1965.
- [37] T.A. Harris, Rolling Bearing Analysis, 3rd ed., John Wiley and Sons, New York, 1991.
- [38] D.L. Bordon, G.A. Holloway, Failure Analysis Gears-Shafts-Bearings-Seals, The Falk Corporation, Milwaukee, WI, 1978.
- [39] A.Z. Szeri (Ed.), Tribology-Friction, Lubrication and Wear, Hemisphere Publication Corp., Mc Graw-Hill Book Co., New York, 1980.
- [40] G. Bernasconi, G. Piatti (Eds.), Creep of Engineering Materials and Structures, Applied Science Publishers, London, 1978.

INDEX

Note: Page numbers followed by *f* indicate figures, and *t* indicate tables.

A

- Abrasive wear, 203
- Acquisition background data, 17
- Adhesive wear, 203–204
- Aero engine
 - bearing cage failure, 148–154
 - bolt of nozzle guide vane failure, 236–240
 - compressor rotor blades failure, 142–147
 - control lever, fuel-metering unit failure, 61–68
 - exhaust outer cone failure, 95–102
 - interrotor bearing failure, 205–214
 - jet nozzle casing failure, 255–258
 - sealing rings, centrifugal breather failure, 102–107
 - turbine stator vanes failure, 259–266
 - vanes of front nozzle diaphragm assembly failure, 240–249
- Aero weapon fin mounting screws failure, 180–183
 - background, 180
 - chemical analysis, 182
 - fractographic examination, 181, 181*f*
 - hardness, 182
 - metallographic examination, 181–182, 181–182*f*
 - visual examination, 180, 180*f*

B

- Banding, 49
- Beach marks, 12–13, 13*f*
- Bearing cage failure, 154
 - background, 148
 - chemical analysis, 153
 - fractographic examination, 150, 150–152*f*
 - hardness, 151
 - metallographic examination, 150–151, 153*f*
 - visual examination, 148, 148–149*f*
- Body of saddle assembly failure, 126–127
 - background, 123–124

- chemical analysis, 126
- fractographic examination, 124, 125–126*f*
- hardness, 124
- metallographic examination, 124, 127–128*f*
- visual examination, 124, 124–125*f*
- Boiler tube, thermal power station failure, 131–132
 - background, 128
 - chemical composition, 131
 - fractographic examination, 128–129, 129*f*
 - hardness, 130–131
 - metallographic examination, 129–130, 130–131*f*
 - visual examination, 128, 129*f*
- Bolts, flapjack-mounting bracket failure, 142
 - background, 139
 - chemical analysis, 142
 - fractographic examination, 140, 140*f*
 - hardness, 141
 - metallographic examination, 141, 141*f*
 - visual examination, 139, 139*f*
- Braggs law, 24
- Breech block, light machine gun failure, 81
 - background, 77
 - chemical analysis, 80
 - fractographic examination, 77, 78–79*f*
 - hardness, 79
 - metallographic examination, 77–79, 79–80*f*
 - recommendation, 81
 - visual examination, 77, 78*f*
- Brinell hardness, 108
- Brittle fracture
 - chevron marks, 9–12, 10*f*
 - coarse/crystalline fracture surface, 7–8
 - embrittlement, types, 8
 - fractography, 8–12
 - herringbone pattern, 9–12, 10*f*

Brittle fracture (*Continued*)

- intergranular fracture, 9–12, 11*f*
- physical factors, 7–8
- river pattern, 9–12
- transgranular cleavage, 9–12

C

Cadmium plating, 76

Cartridge seat ejection secondary (aircraft)
failure

- background, 192–193
- chemical analysis, 200
- fractographic examination, 194,
195–197*f*
- hardness, 199–200
- metallographic examination, 198–199,
198–200*f*
- recommendations, 202
- residual stress measurement, 200
- visual examination, 193–194, 193–194*f*

Casting defects

- blow holes and porosity, 93
- case studies
 - exhaust outer cone failure, 95–102
 - grinding roll failure, 107–110
 - sealing rings, centrifugal breather
failure, 102–107

hot tears, 95

inclusions, 95

pipe, 94

segregation

- chemical inhomogeneity, 93
- electron probe microanalysis
(EPMA), 93
- macro etching, 93

shrinkage, 94

sound foundry practice, 93

Center main bearing (aircraft engine) failure

- background, 249
- chemical analysis, 254
- fractographic examination, 250,
250–251*f*
- hardness, 254
- metallographic examination, 251–253,
251–254*f*
- visual examination, 249, 249*f*

Chemical analysis, 4–5

aeroengine coupling shaft/bevel gear
failure, 232–233

aeroengine interrotor bearing
failure, 212

aero weapon fin mounting screws failure,
182

bearing cage failure, 153

body of saddle assembly failure, 126

bolts, flapjack-mounting bracket failure,
142

breech block, light machine gun failure,
80

cartridge seat ejection secondary (aircraft)
failure, 200

center main bearing (aircraft engine)
failure, 254

combustion chamber outer casing failure,
90

compressor rotor blades failure, 147

control lever, fuel-metering unit failure,
66

drill rods failure, 72

exhaust outer cone failure, 99

fin mounting nuts failure, 47

firing pin, 120mm low range mortar, 36

front nozzle diaphragm assembly vanes
(aero engine) failure, 246–248

fuel metering unit differential valve cover
failure, 185

fuel pipe line failure, 161

gearbox shaft failure, 39

grenade detonator failure, 175

grinding roll failure, 108

high pressure air failure, 169

hip-joint prosthesis failure, 32

hydraulic pipes failure, 169

instrumental analysis, 20

jet nozzle casing (aero engine) failure,
257–258

low temperature super heater (LTSH)
tube failure, 216

main landing gear assembly (aircraft)
failure, 226–227, 228*t*

main under carriage lever assembly
(aircraft) failure, 190–191

mounting bolt, emergency changeover
valve failure, 115

- nozzle guide vane bolt (aero engine)
 - failure, 239
- sample preparation, 20
- sealing rings, centrifugal breather failure, 106
- steam pipe, unit-VI thermal power station failure, 121
- thermal power station condenser tubes failure, 164
- titanium sponge cutting ring failure, 85
- track pins, track link failure, 55
- turbine stator vanes (aero engine) failure, 262
- wet analysis, 20
- wheel hub, aircraft failure, 137
- Chevron marks, 9–12, 10f
- Chloride ions, 156
- Clamshell marks, 12–13
- Combustion chamber outer casing failure, 91
 - background, 86–87
 - chemical analysis, 90
 - fractographic examination, 87, 88–89f
 - hardness, 89–90
 - metallographic examination, 87–88, 90f
 - visual examination, 87, 87f
- Compressor rotor blades failure, 147
 - background, 142
 - chemical analysis, 147
 - fractographic examination, 143–144, 143–146f
 - hardness, 146
 - metallographic examination, 145–146, 146–147f
 - visual examination, 142–143, 143f
- Control lever, fuel-metering unit failure, 67
 - chemical analysis, 66
 - fractographic examination, 62, 63–64f
 - hardness, 65–66
 - metallographic examination, 62–63, 65f
 - overview, 61–62
 - visual examination, 62, 62f
- Corrosion
 - case studies
 - fuel pipe line failure, 157–162
 - grenade detonator failure, 169–178
 - high pressure air pipes failure, 166–169
 - hydraulic pipes failure, 166–169
 - thermal power station condenser tubes failure, 162–166
 - crevice corrosion, 156
 - defined, 155
 - intergranular corrosion, 156
 - pitting corrosion, 155
 - uniform corrosion, 155
- Corrosion debris, 18
- Corrosion-resistant metal, 155
- Coupling shaft/bevel gear (aeroengine) failure
 - background, 228
 - chemical analysis, 232–233
 - chemical composition, 228, 228t
 - hardness, 231
 - mechanical properties, 228, 229t
 - metallographic examination, 229–230, 230–232f
 - visual examination, 229, 229–230f
- Cracking
 - and discontinuities, 19
 - hydrogen pick up, 51
 - under-bead cracking, 51
- Crack propagation, 12
- Crevice corrosion, 156
- D**
- Deeper etching, 22
- Deoxidants, 50–51
- Digital camera, 18
- Dimples
 - elongated dimples, 8–9, 10f
 - equiaxed dimples, 8–9, 9f
- Drill rods failure
 - background, 68
 - chemical analysis, 72
 - fractographic examination, 68, 69–70f
 - hardness, 71–72
 - metallographic examination, 70–71, 71–72f
 - recommendation, 73
 - visual examination, 68, 68–69f
- Ductile fracture, 41–42, 42f
- cup/cone fracture, tensile test, 7, 8f
- fractography, 8–12
- Dye-penetrant test, 19

E

EDAX. *See* Energy dispersive analysis by X-ray (EDAX)

Electron probe micro analysis (EPMA), 23–24, 93, 172, 176*f*

Elevated temperature failures

aluminide-type coatings, 235

case studies

center main bearing (aircraft engine)

failure, 249–255

front nozzle diaphragm assembly vanes

(aero engine) failure, 240–249

jet nozzle casing (aero engine) failure,

255–258

nozzle guide vane bolt (aero engine)

failure, 236–240

turbine stator vanes (aero engine)

failure, 259–266

creep, 235

grain boundary, 235

hot corrosion, 235

thermal fatigue, 235–236

Elongated dimples, 8–9, 10*f*

Energy dispersive analysis by X-ray (EDAX), 21–22, 96–98, 98*f*, 135

EPMA. *See* Electron probe micro analysis (EPMA)

Equiaxed dimples, 8–9, 9*f*

Erosive wear, 203

Etching

deeper etching, 22

macro etching, 93

Exhaust outer cone failure

background, 95

chemical analysis, 99

fractographic examination, 96–98, 97–99*f*

hardness, 99

metallographic examination, 99, 100*f*

recommendations, 101–102

UNS-J92710/ACI-CF8C, 100–101

visual examination, 95–96, 96*f*

F

Fabrication operations, 49

Fatigue failures

case studies

bearing cage failure, 148–154

bolts, flapjack-mounting bracket failure, 139–142

compressor rotor blades failure, 142–147

wheel hub, aircraft failure, 134–139

defined, 133

endurance limit, 133

S-N curve, 133

stages, 133

stress concentration, 133

Fatigue fracture

beach marks, 12–13, 13*f*

clamshell marks, 12–13

defined, 12–13

fractography

fatigue striations, 14, 14*f*

ill-defined fatigue striations, 14, 15*f*

striations with corrosion products, 14, 15*f*

striations with porosity, 14, 16*f*

multiple origins, 12–13, 13*f*

smooth fracture surface, 12–13

Fatigue striations, 14, 14*f*

Ferrous metals, 50

Filler rods, 52

Fin mounting nuts failure

background, 45

chemical analysis, 47

fractographic examination, 45, 46*f*

hardness, 46

metallographic examination, 45–46, 47–48*f*

visual examination, 45, 46*f*

Firing pin, 120mm low range mortar

background, 34–35

chemical analysis, 36

failed firing pin, as-received condition, 35, 35*f*

hardness, 36

maraging steel, 36

metallographic examination, 35–36, 35–36*f*

remedy, 37

Fractographic examination

aeroengine interrotor bearing failure, 206, 207*f*

- aero weapon fin mounting screws failure, 181, 181*f*
 - bearing cage failure, 150, 150–152*f*
 - body of saddle assembly failure, 124, 125–126*f*
 - bolts, flapjack-mounting bracket failure, 140, 140*f*
 - breech block, light machine gun failure, 77, 78–79*f*
 - cartridge seat ejection secondary (aircraft) failure, 194, 195–197*f*
 - center main bearing (aircraft engine) failure, 250, 250–251*f*
 - combustion chamber outer casing failure, 87, 88–89*f*
 - compressor rotor blades failure, 143–144, 143–146*f*
 - control lever, fuel-metering unit failure, 62, 63–64*f*
 - drill rods failure, 68, 69–70*f*
 - exhaust outer cone failure, 96–98, 97–99*f*
 - fin mounting nuts failure, 45, 46*f*
 - front nozzle diaphragm assembly vanes (aero engine) failure, 241–242, 242*f*
 - fuel metering unit differential valve cover failure, 184, 185*f*
 - fuel pipe line failure, 158, 158–160*f*
 - gearbox shaft failure, 38, 38*f*
 - hip-joint prosthesis failure, 30, 32*f*
 - lifting lugs missile failure, 74, 75–76*f*
 - low temperature super heater (LTSH) tube failure, 216, 216*f*
 - main landing gear assembly (aircraft) failure, 220, 221*f*
 - main under carriage lever assembly (aircraft) failure, 188, 189*f*
 - 7.62 mm cartridge cases failure, 41–42, 42–43*f*
 - mounting bolt, emergency changeover valve failure, 113, 113–114*f*
 - nozzle guide vane bolt (aero engine) failure, 236, 237*f*
 - sealing rings, centrifugal breather failure, 102, 103*f*
 - steam pipe, unit-VI thermal power station failure, 119, 119*f*
 - titanium sponge cutting ring failure, 83, 84*f*
 - track pins, track link failure, 53, 54*f*
 - turbine stator vanes (aero engine) failure, 259, 261–262*f*
 - wheel hub, aircraft failure, 135, 136–138*f*
- Fractures
- brittle fracture, 7–8
 - crack formation, 7
 - defined, 7
 - ductile fracture, 7–8
 - fatigue fracture, 12–13
- Front nozzle diaphragm assembly vanes (aero engine) failure
- background, 240
 - carbides, 246–248
 - chemical analysis, 246–248
 - fractographic examination, 241–242, 242*f*
 - hardness, 243–245
 - metallographic examination, 242, 243–248*f*
 - visual examination, 240–241, 241*f*
- Fuel metering unit differential valve cover failure
- background, 183
 - chemical composition, 185
 - fractographic examination, 184, 185*f*
 - hardness, 184
 - metallographic examination, 184, 185–186*f*
 - titanium alloys, 186
 - visual examination, 183–184, 184*f*
- Fuel pipe line failure, 157–162
- background, 157
 - chemical analysis, 161
 - fractographic examination, 158, 158–160*f*
 - hardness, 160
 - metallographic examination, 159–160, 160–161*f*
 - visual examination, 157, 158*f*
- G**
- Gearbox shaft failure, 40
- background, 37
 - chemical analysis, 39
 - failed shaft, as-received condition, 37, 38*f*
 - fractographic examination, 38, 38*f*

Gearbox shaft failure (*Continued*)
 hardness, 39
 metallographic examination, 38, 39–40f
 remedial measures, 40

Grenade detonator failure
 background, 169–170
 chemical analysis, 175
 EDS analysis, 172–173, 177f
 electron probe micro analysis (EPMA),
 172, 176f
 hardness, 173–175
 metallographic examination, 170–172,
 171–175f
 microstructural examination, 177–178
 visual examination, 170, 171f

Grinding roll failure
 background, 107
 chemical analysis, 108
 hardness, 108
 metallographic examination, 107,
 108–109f
 Ni-hard grinding roll, 109–110
 recommendation, 110
 visual examination, 107, 108f

H

Hardness

- aeroengine coupling shaft/bevel gear
 failure, 231
- aeroengine interrotor bearing failure,
 211–212
- aero weapon fin mounting screws failure,
 182
- bearing cage failure, 151
- body of saddle assembly failure, 124
- bolts, flapjack-mounting bracket failure,
 141
- breech block, light machine gun failure,
 79
- cartridge seat ejection secondary (aircraft)
 failure, 199–200
- center main bearing (aircraft engine)
 failure, 254
- checking, 4–5
- combustion chamber outer casing failure,
 89–90
- compressor rotor blades failure, 146

- control lever, fuel-metering unit failure,
 65–66
- drill rods failure, 71–72
- exhaust outer cone failure, 99
- fin mounting nuts failure, 46
- firing pin, 120mm low range mortar, 36
- front nozzle diaphragm assembly vanes
 (aero engine) failure, 243–245
- fuel metering unit differential valve cover
 failure, 184
- fuel pipe line failure, 160
- gearbox shaft failure, 39
- grenade detonator failure, 173–175
- grinding roll failure, 108
- hip-joint prosthesis failure, 32
- hydraulic pipes failure, 167–168
- improper heat treatment, 25
- jet nozzle casing (aero engine) failure,
 256–257
- lifting lugs missile failure, 75
- low temperature super heater (LTSH)
 tube failure, 216
- main landing gear assembly (aircraft)
 failure, 225
- main under carriage lever assembly
 (aircraft) failure, 190
- microstructural deficiency, 25
- 7.62 mm cartridge cases failure, 43
- mounting bolt, emergency changeover
 valve failure, 115
- nozzle guide vane bolt (aero engine)
 failure, 239
- Rockwell hardness, 24
- sealing rings, centrifugal breather failure, 106
- steam pipe, unit-VI thermal power station
 failure, 121–122
- thermal power station condenser tubes
 failure, 163
- titanium sponge cutting ring failure, 85
- track pins, track link failure, 53
- turbine stator vanes (aero engine) failure,
 261–262
- wheel hub, aircraft failure, 135

Heat treatment, 49

- improper heat treatment, 50
- quench cracking, 50

Herringbone pattern, 9–12, 10f

High pressure air pipes failure
background, 166
chemical analysis, 169
metallographic examination, 166–167,
167–168*f*
visual examination, 166, 166*f*

Hip-joint prosthesis failure
background, 30
boiling nitric acid test, 33–34
chemical analysis, 32
fractographic examination, 30, 32*f*
hardness, 32
metallographic examination, 30, 32–33*f*
nonmetallic inclusions, 33–34
remedial measures, 34
visual examination, 30, 31*f*

Hot tears, 95

Human error, 3

Hydraulic pipes failure
background, 166
chemical analysis, 169
hardness, 167–168
metallographic examination, 166–167,
167–168*f*
visual examination, 166, 166*f*

Hydrogen embrittlement
case studies
aero weapon fin mounting screws
failure, 180–183
cartridge seat ejection secondary
(aircraft) failure, 192–202
fuel metering unit differential valve
cover failure, 183–187
main under carriage lever assembly
(aircraft) failure, 187–192
defined, 179
stress corrosion cracking, 179–180

I

Ill-defined fatigue striations, 14, 15*f*

Impact toughness, 26

Imperfections
and discontinuities, 93
in materials, 29

Inclusions, 95

Instrumental analysis, 20

Intergranular corrosion, 156–157

Intergranular fracture, 9–12, 11*f*

Interrotor bearing (aeroengine) failure
background, 205
cage, 209, 212–213*f*
chemical analysis, 212
fatigue cracks, 213
fractographic examination, 206, 207*f*
hardness, 211–212
inner race, 208, 210–211*f*
metallographic examination, 207
outer race, 207–208, 208–209*f*
rollers, 208–209, 211–212*f*
visual examination, 205, 205–207*f*

J

Jet nozzle casing (aero engine) failure
background, 255
chemical analysis, 257–258
experimental results, 255–257
hardness, 256–257
hot corrosion, 257*f*, 258
metallographic examination, 256,
256–257*f*
visual examination, 255–256, 255*f*

L

Lifting lugs missile failure
background, 73–74
cadmium plating, 76
chemical composition, 75
fractographic examination, 74, 75–76*f*
hardness, 75
metallographic examination, 74–75
visual examination, 74, 74–75*f*

Low-power microscope (stereo
microscope), 19

Low temperature super heater (LTSH) tube
failure
background, 214
chemical analysis, 216
fractographic examination, 216, 216*f*
hardness, 216
metallographic examination, 216, 217*f*
surface erosion, 216–218
visual examination, 214–216, 215*f*

M

- Macro etching, 93
- Main landing gear assembly (aircraft) failure
 - background, 218
 - chemical analysis, 226–227, 228*t*
 - fractographic examination, 220, 221*f*
 - frictional overheating, 228
 - hardness, 225
 - metallographic examination, 220–222, 222–226*f*
 - visual examination, 218–220, 219–220*f*
- Main under carriage lever assembly (aircraft) failure
 - background, 187
 - chemical analysis, 190–191
 - fractographic examination, 188, 189*f*
 - hardness, 190
 - metallographic examination, 188–190, 190–191*f*
 - visual examination, 187–188, 187–188*f*
- Manufacturing defects
 - banding, 49
 - carbon steels, 50–51
 - case studies
 - breech block, light machine gun failure, 77–81
 - combustion chamber outer casing failure, 86–91
 - control lever, fuel-metering unit failure, 61–68
 - drill rods failure, 68–73
 - lifting lugs missile failure, 73–77
 - titanium sponge cutting ring failure, 82–86
 - track pins of track link failure, 52–56
 - track shoe infantry combat vehicle failure, 56–61
 - chemical segregation, 49
 - cracking, 51
 - deoxidants, 50–51
 - fabrication operations, 49
 - ferrous metals, 50
 - fusion welding process, 51
 - heat treatment, 49
 - improper heat treatment, 50
 - quench cracking, 50
 - nickel base alloys, 52
 - nonferrous metals, 51–52
 - nonheat treatable aluminum alloys, 52
 - residual stresses, 49
 - subsurface/core discontinuities, 49
 - surface discontinuities, 49
 - welding, 50
- Material selections
 - case studies
 - fin mounting nuts failure, 45–48
 - firing pin, 120mm low range mortar, 34–37
 - gearbox shaft failure, 37–41
 - hip-joint prosthesis failure, 30–34
 - 7.62 mm cartridge cases failure, 41–45
 - factors, 29
 - imperfections, 29
 - special materials, 29
 - structural materials, 29
- Metallographic examination
 - aeroengine coupling shaft/bevel gear failure, 229–230, 230–232*f*
 - aero weapon fin mounting screws failure, 181–182, 181–182*f*
 - bearing cage failure, 150–151, 153*f*
 - body of saddle assembly failure, 124, 127–128*f*
 - bolts, flapjack-mounting bracket failure, 141, 141*f*
 - breech block, light machine gun failure, 77–79, 79–80*f*
 - cartridge seat ejection secondary (aircraft) failure, 198–199, 198–200*f*
 - center main bearing (aircraft engine) failure, 251–253, 251–254*f*
 - combustion chamber outer casing failure, 87–88, 90*f*
 - compressor rotor blades failure, 145–146, 146–147*f*
 - control lever, fuel-metering unit failure, 62–63, 65*f*
 - drill rods failure, 70–71, 71–72*f*
 - exhaust outer cone failure, 99, 100*f*
 - fin mounting nuts failure, 45–46, 47–48*f*
 - firing pin, 120mm low range mortar, 35–36, 35–36*f*
 - front nozzle diaphragm assembly vanes (aero engine) failure, 242, 243–248*f*

fuel metering unit differential valve cover failure, 184, 185–186*f*
fuel pipe line failure, 159–160, 160–161*f*
gearbox shaft failure, 38, 39–40*f*
grenade detonator failure, 170–172, 171–175*f*
grinding roll failure, 107, 108–109*f*
high pressure air pipes failure, 166–167, 167–168*f*
hip-joint prosthesis failure, 30, 32–33*f*
hydraulic pipes failure, 166–167, 167–168*f*
jet nozzle casing (aero engine) failure, 256, 256–257*f*
lifting lugs missile failure, 74–75
low temperature super heater (LTSH) tube failure, 216, 217*f*
main landing gear assembly (aircraft) failure, 220–222, 222–226*f*
main under carriage lever assembly (aircraft) failure, 188–190, 190–191*f*
7.62 mm cartridge cases failure, 42–43, 43–44*f*
mounting bolt, emergency changeover valve failure, 115, 115–116*f*
nozzle guide vane bolt (aero engine) failure, 238–239, 238–239*f*
sealing rings, centrifugal breather failure, 104–105, 104–105*f*
steam pipe, unit-VI thermal power station failure, 119–120, 120–121*f*
thermal power station condenser tubes failure, 162, 163–165*f*
titanium sponge cutting ring failure, 84, 85*f*
track pins, track link failure, 53, 54–55*f*
turbine stator vanes (aero engine) failure, 259, 263–265*f*
wheel hub, aircraft failure, 135, 138*f*
Microscopy, 22
7.62 mm cartridge cases failure, 45
background, 41
fractographic examination, 41–42, 42–43*f*
hardness, 43
metallographic examination, 42–43, 43–44*f*
visual examination, 41, 41*f*

Mounting bolt, emergency changeover valve failure, 116–117
background, 112, 112*f*
chemical analysis, 115
fractographic examination, 113, 113–114*f*
hardness, 115
metallographic examination, 115, 115–116*f*
visual examination, 113, 113*f*

N

Nickel base alloys, 52
Nitrogen, 95
Ni-hard grinding roll, 109–110
Nondestructive testing (NDT), 19
Nonferrous metals, 51–52
Nonheat treatable aluminum alloys, 52
Nonmetallic inclusions, 33–34
Nozzle guide vane bolt (aero engine) failure
background, 236
carbides, 239–240
chemical analysis, 239
fractographic examination, 236, 237*f*
hardness, 239
metallographic examination, 238–239, 238–239*f*
visual examination, 236, 236*f*

O

Optical microscopy, 22–23
Overload failures
case studies
body of saddle assembly failure, 123–127
boiler tube, thermal power station failure, 128–132
mounting bolt, emergency changeover valve failure, 112–117
steam pipe, unit-VI thermal power station failure, 117–123
ductile/brittle mode, 111
factor of safety, 111
metallurgical defects, 111
Oxidizing metal ions, 155

P

Phase analysis, 24
Photographic records, 18

Pipe, 94
Pitting corrosion, 155
Plastic deformation, 8–9
Preliminary examination, 18–19
Primary pipe, 94

Q

Quasi-cleavage fracture, 12, 12*f*
Quench cracking, 50

R

Radiography, 19
 of castings, 94
 primary pipe, 94
Residual stresses, 49
River pattern, 9–12
Rockwell hardness, 24

S

Samples selection, 17
Scanning electron microscope (SEM), 87, 117–118, 124, 135
 dimples, 8–9
 fatigue striations, 14, 14*f*
 ill-defined fatigue striations, 14, 15*f*
 intergranular fracture, 9–12, 11*f*
 principle of, 21
 quasi-cleavage fracture, 12, 12*f*
 striations with corrosion products, 14, 15*f*
 striations with porosity, 14, 16*f*
 transgranular cleavage fracture, 9–12, 11*f*
Sealing rings, centrifugal breather failure, 106
 background, 102
 chemical analysis, 106
 fractographic examination, 102, 103*f*
 hardness, 106
 metallographic examination, 104–105, 104–105*f*
 visual examination, 102, 102*f*
Secondary electron image, 21
SEM. *See* Scanning electron microscope (SEM)
Shredder system, 82
Shrinkage, 94
Shrinkage cavity, 94
Solute rejection, 93

Steam pipe, unit-VI thermal power station failure, 122–123
 background, 117
 chemical analysis, 121
 fractographic examination, 119, 119*f*
 hardness, 121–122
 mechanical properties, 121–122
 metallographic examination, 119–120, 120–121*f*
 stress calculation, 122
 visual examination, 117–118, 118–119*f*
Stereomicroscope, 19
Stress corrosion cracking, 179–180
Subsurface/core discontinuities, 49
Sulfides, 95
Surface discontinuities, 49

T

Tempered martensite embrittlement (TME), 85–86
Tensile properties, 25
Thermal power station condenser tubes failure
 background, 162
 chemical analysis, 164
 hardness, 163
 metallographic examination, 162, 163–165*f*
 visual examination, 162, 162*f*
Titanium alloys, 186
Titanium sponge cutting ring failure
 background, 82
 chemical analysis, 85
 fractographic examination, 83, 84*f*
 hardness, 85
 metallographic examination, 84, 85*f*
 remedial measures, 86
 tempered martensite embrittlement (TME), 85–86
 visual examination, 82–83, 82–83*f*
Track pins, track link failure, 55
 background, 52–53
 chemical analysis, 55
 fractographic examination, 53, 54*f*
 hardness, 53
 metallographic examination, 53, 54–55*f*
 recommendations, 56
 visual examination, 53, 53–54*f*

- Track shoe infantry combat vehicle failure, 61
 - background, 56
 - chemical analysis, 60–61
 - fractographic examination, 57, 57–59*f*
 - hardness, 59
 - metallographic examination, 58–59, 59–60*f*
 - visual examination, 56–57, 56–57*f*
- Transgranular cleavage, 9–12, 11*f*
- Turbine stator vanes (aero engine) failure
 - aerofoil region cracking, 265–266
 - age hardening temperature, 262–263
 - background, 259
 - chemical analysis, 262
 - fractographic examination, 259, 261–262*f*
 - hardness, 261–262
 - high-temperature cobalt base alloys, 262–263
 - interdendritic oxide-filled surface cracks, 264–265
 - metallographic examination, 259, 263–265*f*
 - visual examination, 259, 259–260*f*

U

- Uniform corrosion, 155

V

- Vickers hardness, 24
- Visual examination, 18
 - aeroengine coupling shaft/bevel gear failure, 229, 229–230*f*
 - aeroengine interrotor bearing failure, 205, 205–207*f*
 - aero weapon fin mounting screws failure, 180, 180*f*
 - bearing cage failure, 148, 148–149*f*
 - body of saddle assembly failure, 124, 124–125*f*
 - bolts, flapjack-mounting bracket failure, 139, 139*f*
 - breech block, light machine gun failure, 77, 78*f*
 - cartridge seat ejection secondary (aircraft) failure, 193–194, 193–194*f*

- center main bearing (aircraft engine) failure, 249, 249*f*
- combustion chamber outer casing failure, 87, 87*f*
- compressor rotor blades failure, 142–143, 143*f*
- control lever, fuel-metering unit failure, 62, 62*f*
- drill rods failure, 68, 68–69*f*
- exhaust outer cone failure, 95–96, 96*f*
- fin mounting nuts failure, 45, 46*f*
- front nozzle diaphragm assembly vanes (aero engine) failure, 240–241, 241*f*
- fuel metering unit differential valve cover failure, 183–184, 184*f*
- fuel pipe line failure, 157, 158*f*
- grenade detonator failure, 170, 171*f*
- grinding roll failure, 107, 108*f*
- high pressure air pipes failure, 166, 166*f*
- hip-joint prosthesis failure, 30, 31*f*
- hydraulic pipes failure, 166, 166*f*
- jet nozzle casing (aero engine) failure, 255–256, 255*f*
- lifting lugs missile failure, 74, 74–75*f*
- low-power microscope (stereo microscope), 19
- low temperature super heater (LTSH) tube failure, 214–216, 215*f*
- main landing gear assembly (aircraft) failure, 218–220, 219–220*f*
- main under carriage lever assembly (aircraft) failure, 187–188, 187–188*f*
- 7.62 mm cartridge cases failure, 41, 41*f*
- mounting bolt, emergency changeover valve failure, 113, 113*f*
- nozzle guide vane bolt (aero engine) failure, 236, 236*f*
- sealing rings, centrifugal breather failure, 102, 102*f*
- steam pipe, unit-VI thermal power station failure, 117–118, 118–119*f*
- thermal power station condenser tubes failure, 162, 162*f*
- titanium sponge cutting ring failure, 82–83, 82–83*f*
- track pins, track link failure, 53, 53–54*f*

Visual examination (*Continued*)

- turbine stator vanes (aero engine) failure, 259, 259–260*f*
- wheel hub, aircraft failure, 134–135, 134–135*f*

W

Wear

- abrasive wear, 203
- adhesive wear, 203–204
- case studies
 - coupling shaft/bevel gear (aeroengine) failure, 228–233
 - interrotor bearing (aeroengine) failure, 205–214
 - low temperature super heater (LTSH) tube failure, 214–218
 - main landing gear assembly (aircraft) failure, 218–228

- effective lubrication, 204
- erosive wear, 203
- ineffective lubrication, 204

Weld decay, 156

Welding, 50

Wet analysis, 20

Wheel hub, aircraft failure, 138–139

- background, 134
- chemical analysis, 137
- fractographic examination, 135, 136–138*f*
- hardness, 135
- metallographic examination, 135, 138*f*
- visual examination, 134–135, 134–135*f*

X

- X-ray diffraction, 24
- X-ray energy spectrum, 21
- X-ray photon, 23