
CROWOOD METALWORKING GUIDES

ENGINEERING MATERIALS

HENRY TINDELL



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Safety is of the utmost importance in every aspect of metalworking. The practical workshop procedures and the tools and equipment used in metalworking are potentially dangerous. Tools should be used in strict accordance with the manufacturer's recommended procedures and current health and safety regulations. The author and publisher cannot accept responsibility for any accident or injury caused by following the advice given in this book.

Photography by Colin Mills

Dedication

To Susan Elizabeth Tindell (1949–2012).

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Preface

In a parallel universe there may exist the perfect material – as strong as an alloy steel, hard as diamond, malleable as gold, corrosion-resistant as titanium, weldable as mild steel, cheap as grey cast iron, abundant as wood, long-lasting as flint, etc. But, alas, it doesn't occur here, hence the need to understand and select from the myriad materials commonly available. There are, for instance, literally thousands of different steels produced worldwide, although we need only know the details of a tiny selection.

Whilst the material in a project usually accounts for only a small part of the cost – and it is then the subject of considerable 'added value' in processing (e.g. machining) – it is surely appropriate to avoid this effort being wasted on the wrong material. Apparently invulnerable items may be rendered useless by the insidious effects of, for instance, vibration (slow crack growth – fatigue); stress-corrosion cracking (austenitic stainless steel in certain environments); or fatally cracked (incorrect welding procedure for a low-alloy steel).

This book should help as a guide through the labyrinth that is engineering material applications, based on well-established theory and practically available materials.

Henry Tindell
Cheadle Hulme, 2013

Introduction

'It's better to burn than rust...', sang the Eagles in 1972 but, like the Rolling Stones, forty years later they've discovered that it's better to rust gently than burn, just like my old bike...

For those of the gently rusting persuasion, this book aims to explain the application of engineering materials, essentially metals, for the engineer who is concerned with making things. The model engineer, the amateur or even professional experimental engineer, really needs a basic feeling for materials otherwise a lot of the enjoyment of these activities will never be realized; not to mention the pitfalls into which the unwary can fall, with expensive failures incurring lost time, money or worse...

In the post-industrial age now upon the West, practical skills and knowledge are at a premium, for not all things can be facilitated directly from the virtual world. And there is, despite the present information revolution, a common lack of basic understanding of the materials used in making things.

Metals have become ubiquitous since the early twentieth century; like computers in our age they are often overlooked. The great advances in understanding of metals' behaviour, with many discoveries as recently as in the last fifty-odd years, may not be apparent to the non-specialist worker. Understanding how metals work has increasingly become the province of the materials specialist, as Further and Higher Education courses have moved on from metallurgy to the 'newer' materials like plastics and composites, with computer modelling rather than hands-on engagement. This is all very well, but it is leaving a great hole for expertise 'in the shed', when a substantial industrial manufacturing facility cannot be called upon for support.

Even a generation ago, it was not uncommon (in the UK, at least) for an engineering firm to be sufficiently vertically integrated that many trades were available in-house, from departments for electrics, plumbing and joinery, to design, fabrication, foundry-work, mechanical and non-destructive testing – even metallurgical analysis. This meant that often substantial domestic jobs could be tackled with access to these trades, like re-wiring a house and shed, or fitting a central heating system. It provided a source of material knowledge, even the materials themselves, for that special model or vehicle repair. (A downside of the latter I once saw was when a lunchtime welding repair on a car ended in it catching alight and being written off – but that's the cost of experience!) Nowadays, this scenario is rare and the enterprising individual has to look elsewhere for help and guidance, with the result that an understanding of materials is becoming an impoverished area of expertise. There is, perhaps, a parallel with cooking, where time and circumstances have tempted a generation away from individual food preparation skills and into fast foods and takeaways. The burgeoning interest in cooking programmes and literature, however, shows the inherent desire for a practical connection with the *process* – the 'making things' drive, not dissimilar to that of the practical engineer.

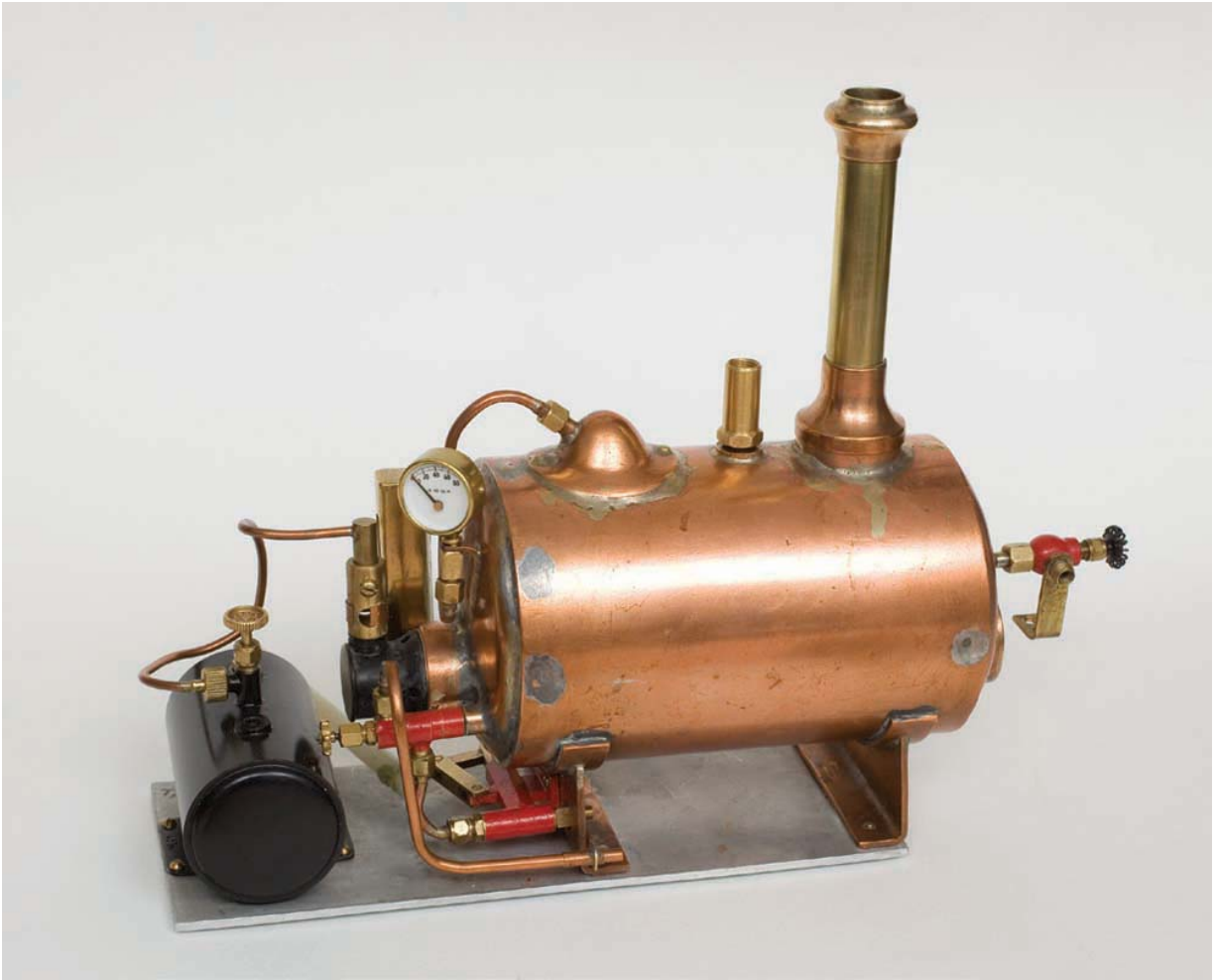


Photo 1 Horizontal boiler, for use with a steam engine in a glass fibre model boat.



Photo 2 Vertical boiler, delivering 30psi, to power a Viper Vee twin, oscillator steam engine.



Photo 3 Early twentieth century vertical steam engine, from a German manufacturer.

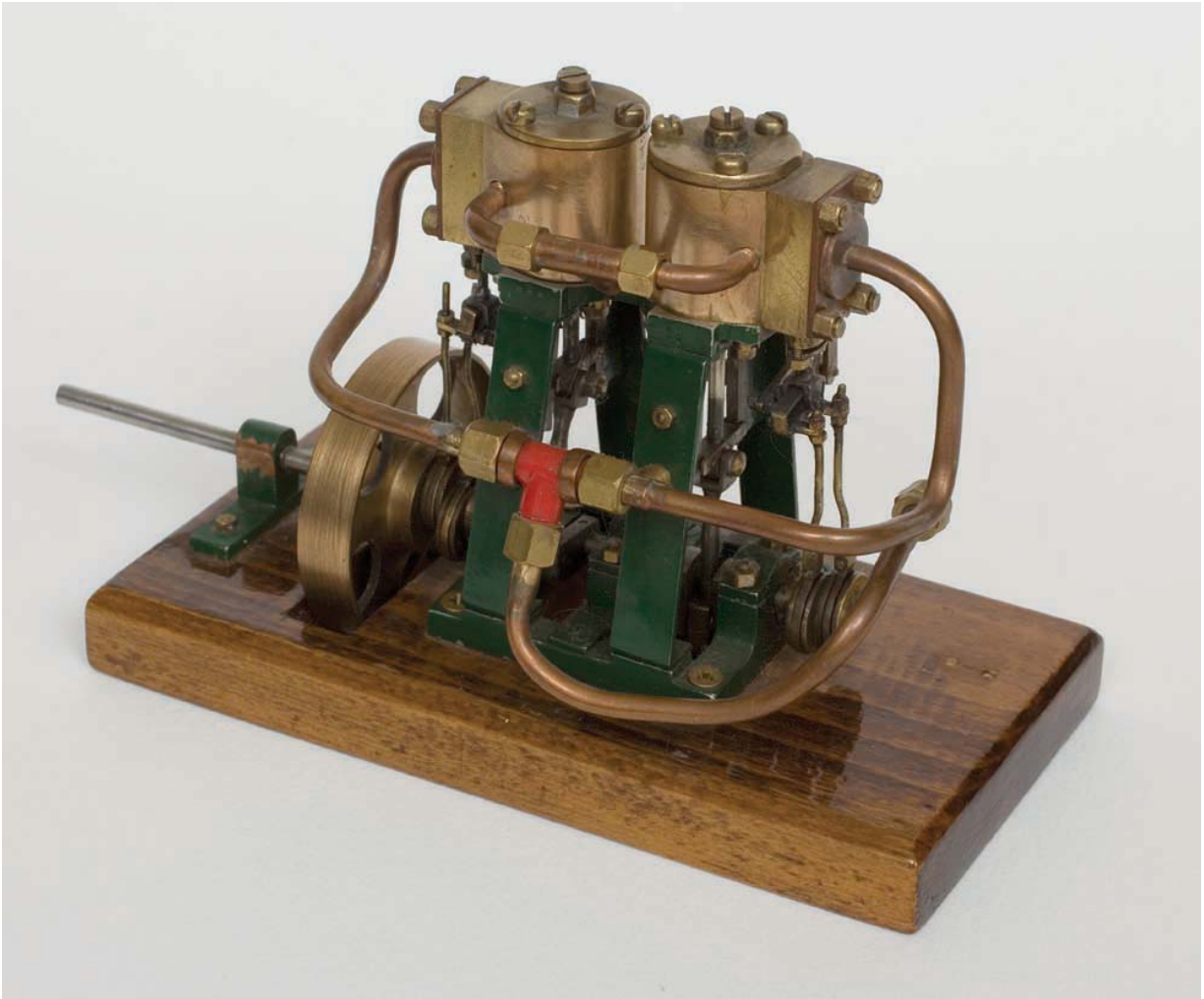


Photo 4 Vertical parallel twin cylinder marine engine, often adapted as factory power source.

In this increasingly ethereal world of the virtual image, the benefit of making real things can be remarkably therapeutic, but in materials-driven engineering projects large or small, information and experience need to be steadily gained or disappointment and failure can overwhelm the enthusiastic worker. So, this book tries to describe the metals commonly available to the general enthusiast and how they can be selected and processed to a satisfactory product. As importantly, it points to the limitations of metals – an area that demands sufficient knowledge and experience.

Whilst considerable attention is normally paid to engineering processes, such as lathe-work, milling, bench-work and welding, a

basic knowledge of the *materials* being worked is largely ignored. Indeed, there is a parallel with professional engineering, where the understanding of metals can surprisingly often be treated as a 'black art'. But there is no good reason not to explore this fascinating field, and in the following chapters the aim is to describe the variety of materials available, and to show how the behaviour of metals can be understood, for their successful application and avoidance of failures.

'Why not just make everything out of mild steel?' is a question that will be explored, basically depending on the project undertaken. Only by understanding aspects like mechanical behaviour and corrosion resistance of the chosen material can we successfully fit it to the task. For instance, whilst mild steel is justifiably popular with the model engineer, it would be folly to employ it in highly stressed parts like vehicle transmissions, weight-sensitive applications in flying, or in corrosive environments.

The photographs in this chapter show some examples of small models connected with steam power, a favourite with model engineers worldwide, made largely from copper-base alloys, such as copper and brass, and mild steel.

A BRIEF HISTORY OF ENGINEERING MATERIALS

The history of materials has largely determined the way of life and progress of civilizations ever since mankind has inhabited the earth – the eponymous ages of Stone, Bronze and Iron are familiar to all, if somewhat misleading.

Instead, the great sweep of materials history can be seen as three distinct periods. The first, from 3500BC, continued until the Industrial Revolution around AD1700. This primarily made use of wood and stone with metals being deployed on only a small scale, despite the implied usage of metals in the Bronze and Iron Ages.

The second age began with the Industrial Revolution and by the early 1800s steam began to challenge water power as the energy source of choice. By 1850 steam power, generated by extensive

coal supplies in the UK, had supplanted water power. Steam engines were driving all forms of industry and enabling the explosive growth of the railways. These developments often required improved materials and knowledge. As this sometimes lagged behind construction, disasters resulted (the collapse of the Tay Bridge, for example) as the shortcomings of designs using non-ductile materials in tension, such as cast iron, were discovered.

The early twentieth century ushered in the third age which continues with the proliferation of truly 'designer' materials. The range of engineering metals is now being challenged by many developments in the synthetic non-metals, like ceramics and composites, which have taken several decades to reach application and are only now being extensively used in mainstream civil aircraft and high-performance sports products; these will continue to grow in the twenty-first century. Application of science to materials in the second age, like low-cost ductile steels, has undergone exponential growth in the third age, where we now understand how cast irons can be made almost as ductile as mild steel, why soft aluminium can be made as strong as the steel which is three times heavier, and how the debilitating effects of corrosion can be avoided – car bodies now lasting far longer than before. Indeed, air transport as we know it would be impossible without the long-term development of titanium and high-nickel alloys for jet engines.

Table 1 Materials timeline.

	Date	Common materials	Application examples	Some manufacturing processes
BC	100,000	wood, bone , flint	tools and weapons	hand carving, hammering, chipping, flaking
BC	10,000	iron meteorites, copper, gold, bronze, glass	jewellery, clothing, statues, tools and weapons	metal ores reduction, kiln firing
BC	3,000	silver, lead	jewellery, vessels	casting, forging, drawing of wire
BC	2,000	iron, steel, tin, glue	kiln bricks, machines from wood, with iron and steel pins	charcoal smelting (iron), cementation process
BC	1,000	Iron Age (Britain), coal (China), parchment (Egypt)	agriculture (ploughs, horseshoes), chains, nails, bellows (leather)	steel hardening by quenching, wood turning lathe, glass moulding
AD	0	gunpowder (China), concrete (Roman), paper	waterwheel mills, glass for windows, bells	steel tempering (Roman)
AD	1000	cast iron (Europe), rubber (S. America)	early cannons, printing instruments	smelting with coal, power from water for sawmills
AD	1500	coke, zinc (India)	rails for moving goods, cast iron weapons	working iron in rolling mills
AD	1800	cobalt, chromium, manganese, molybdenum, nickel	steam engines for power, early machine tools	cast iron and steel production, factories start
AD	1900	aluminium, magnesium, many other elements, low cost steel, alloy steels, industrial diamond, dynamite	steam engine development, Railway Age, internal combustion engine, electric power, etc.	low cost plain and alloy steels, structural aluminium, vulcanized tyres, exponential growth of manufacturing processes and scientific understanding of

EARLY MATERIALS

Even as many as 250,000 years ago, materials were being deployed for hunting and, much later, agriculture. These were limited to the readily available stone (flint), animal bones, hair and skin, and wood from local forests. Sharpened flints provided powerful tools for the Stone Age, making hammers, saws, knives, needles, hooks and even arrow-heads for the wooden bows in use by 20,000BC. Ceramics (small items of glazed pottery) were available by 12,000BC. Even ornamental glass was being made in Egypt by 7,000BC.

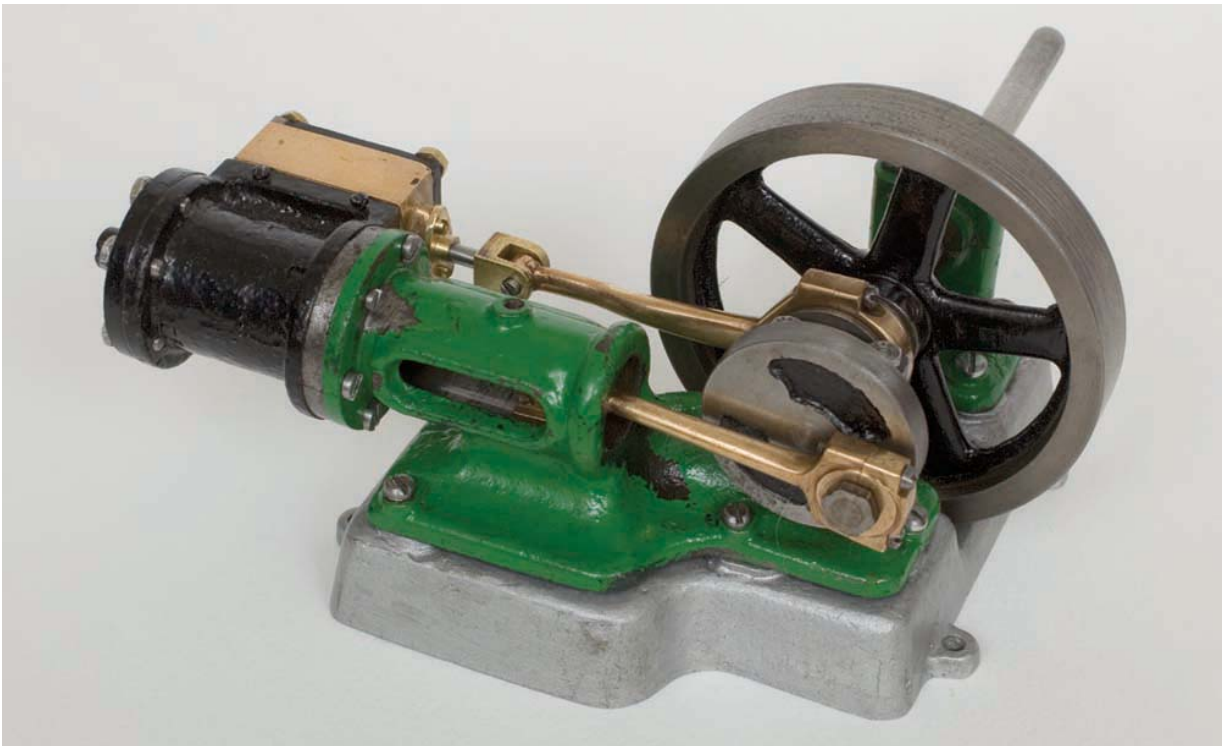


Photo 5 Horizontal single cylinder steam engine, manufactured c. 1960, power source for small manufactory or workshop.

Early metals

The Middle East is generally regarded as the birthplace of the systematic use of metals and by 3,500BC the Sumerians, having already developed wooden wheels and axles, were casting, forging and forge-welding copper from local ores. Not until 2,000BC did the knowledge of bronze lead to its eponymous Age, with another 500 years before the widespread working of ferrous ores began the Iron Age. These were not, however, distinct periods, one supplanting the other like some latter-day vinyl/tape/CD/ANO development – rather the techniques were selected or co-existed over millennia, sometimes even into the modern era.



Photo 6 General purpose proprietary vertical engine, popular in factory mills and workshops, seen in the film The African Queen.

Stone

Stone buildings were being developed around the Mediterranean, in the Middle East and North Africa from 3,500BC – famously from 2,700BC for Egypt's temples and pyramids. The use of 1,000-tonne blocks was a testament to their highly developed engineering in wood for the manipulation of these enormous pieces. This knowledge slowly spread across the western world, to the England of Stonehenge, 1,000BC. However, whilst one cannot but be impressed by the ancient Egyptians' enterprises, it shouldn't be forgotten that practices were often more 'trial-and-error' than scientific theory – witness the catastrophic collapse of the Fourth Dynasty King Sneferu's pyramid at Mendum during, or just after, construction.



Photo 7 Simple beam engine as used in the early nineteenth century Industrial Revolution for general power source.

Wood

It has, fairly, been said that in the history of materials, wood has been the mild steel of every age until the latter part of the Industrial Revolution. Its properties are impressive (with careful selection), providing strength in tension, compression and bending, with predictable behaviour, light weight, ready availability and ease of shaping. The experience of centuries meant that it was not until the demands of industrial production and the factory system, plus the ever-present desire for advantage in warfare, finally led to metals largely superseding wood as the pre-eminent engineering material.



Photo 8 'Mamod Minor' horizontal steam engine, manufactured late 1930s to 1940s.

EARLY METAL DEVELOPMENT

The development of kiln-dried bricks in Assyria in around 1,000BC enabled primitive hearths to be used for bronze, and then iron smelting at significantly higher temperatures. Perhaps the most important driver for metals was the rise and fall of the various empires around central Europe, and the deployment of their large armies. By 500BC the Egyptian empire was overtaken by the Persians, then Alexander the Great established the Greek supremacy which subsequently fell to the Romans who ruled from North Africa to Britain from 250BC to AD400.

This led to steady developments in metals, primarily for the military with the Greeks beginning to understand the rudiments of hardening by quenching of their iron (laboriously made with enough diffused carbon to permit hardening). The Romans progressed to tempering the quenched iron-carbon alloy, in their quest for the best combination of strength and toughness for critical weaponry like their famous short swords. Records exist that show how these swords helped to defeat the Celts, with their inferior swords bending on impact in the desperate struggle of pitched battle.



Photo 9 'ESL' steam power unit of 1950s, popular for Meccano model operation.

Developments in the Far East

Long before the Greeks and Romans in the West, the early developments in metals technology split eastwards to India and China. This produced separate development, and superior advances in China with gunpowder, cast iron and high-calibre composite steels, a thousand years in advance of the West. This advantage, however, was lost with the chronic dislocation in communication and trade with the West until modern times when this is being redressed

in dramatic fashion. While the West has lost much industrial power to the East, the amateur has at least benefited from the familiar high-quality, with low cost, tooling.

Early mechanics

Alongside their materials progress, the ancient Greeks developed an understanding of the elements of machines, like the wheel, lever, pulley and screw, but their mechanisms failed to make significant use of metal – one important limitation being the difficulty in working with hard materials, compared to the relative ease of working in wood.

We have records from Philo of Byzantium (2,000BC) which describe early experiments in elasticity, testing deflection of swords and bronze catapult springs under controlled loading. The Romans, as great miners, also made much use of lead and bronze for domestic as well as military use.

Materials in the Dark Ages

After the fall of the Roman Empire, the Dark Ages in Europe saw a return to stone and wood as primary materials, although some gradual improvement in machinery, such as waterwheels and quern mills, was made – albeit without significant deployment of metal. The two areas demanding use of metal instead of wood were armaments and clock-making. As the world's first automatic machines, clocks, then watches, required all-metal construction as horology developed in the thirteenth and fourteenth century.

Materials in the Renaissance

In the fifteenth century the invention of printing, with moveable type, was enabled by fine casting methods, although lost-wax casting had been successfully developed in China a thousand years previously. This first Information Revolution presaged the Renaissance in

Europe, with Leonardo da Vinci (1452–1519) undertaking studies into the properties of wires and beams. His simple tensile testing machine for wires was the principle used when the author worked at Fulmer Research Institute, Stoke-Poges, England, in the 1970s. We were studying the behaviour, for patent applications, of copper-base alloys ('brasses') exhibiting 'memory' properties associated with unusual strain-induced martensitic transformations. The device used a hopper which allowed lead shot to fall into a container attached to the specimen, thus loading it as desired for a special tensile test. Extremely simple and effective, it would surely have amused the great Leonardo!

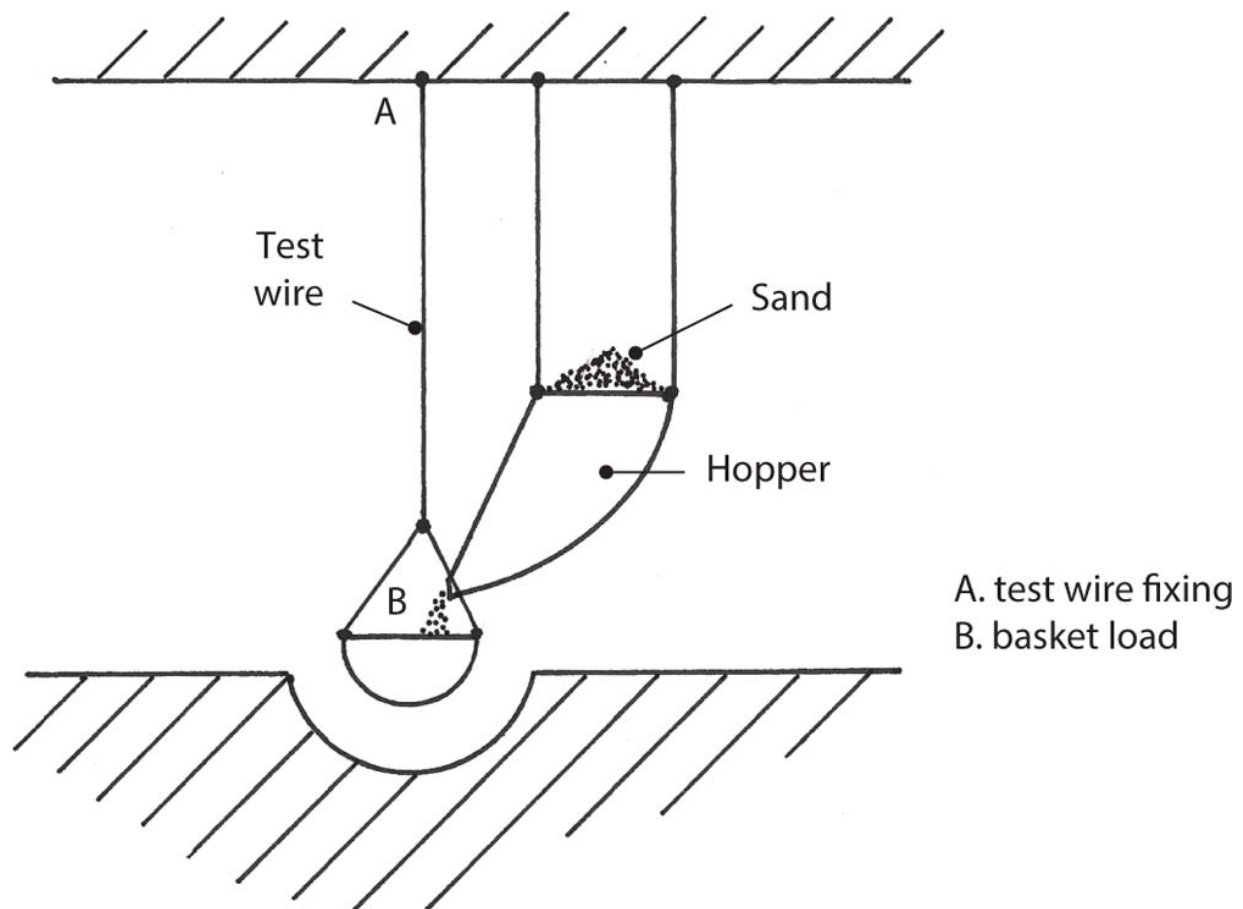


Figure 1 Tensile testing – Leonardo da Vinci's apparatus.

Developments in the strength of materials

In his book *Two New Sciences* (1638), Galileo of Florence was the first to discuss 'stress analysis', a pioneering work on strength of materials. He even came close to correctly understanding the bending of beams theory which underpins structural analysis. In England, several decades later, Robert Hooke provided a framework for elasticity and the relation between stress and strain (load versus extension) that is at the heart of predicting a material's behaviour. Hooke's Law simply shows how the elastic (i.e. *reversible*) behaviour of metals can be confidently predicted from their unique stress-strain graph, upon which all stress analysis is based, as seen in [Chapter 3](#).

There followed considerable efforts by the French theorists in the seventeenth and eighteenth centuries, such as Mariotte (internally pressurized pipes; laterally loaded plates) and Parent, who finally cracked the beam bending theory in 1713. Important theories from Poisson, Lamé, Clapyron and St Venant of the École Polytechnique in Paris between the late eighteenth and late nineteenth centuries provided comprehensive analysis of beams, plates, shells and structures, laying the foundations of modern structural analysis and the pressure vessel codes that finally overcame the problem of harnessing high-pressure steam. This had dogged the use of steam power through the Industrial Revolution, leading to many unexpected and often catastrophic boiler explosions and the rise of the industrial insurance industry.

By the end of the nineteenth century, studies in the strength of materials were sufficiently advanced for Giraud to produce his work on the subject, building on the work of Coulomb (1736–1806) who explored scientifically the effect of heat treatment upon the strength of iron. Whilst the French approached the subject theoretically, the experimental work characterized by the British approach provided significant practical advances. The rise of industry in the USA contributed rapid advances in machining processes with Capstan and automatic lathes, as well as steel structures with ever-larger bridges and buildings. In Germany, Wöhler started vital work on rotating shafts failing by fatigue, and Neumann (1798–1895)

developed visualization of stress with photo-elasticity. This built on the work in England of Thomas Young, 1773–1829 (of *Young's Modulus*), founder of the Cambridge School of Materials. In St Petersburg an important Engineering School was established in the early nineteenth century, with prominent contributions from the French theoreticians and important Russian contributions to modern theory on shear stresses in beams, in current use.

Material limitations

As the Railway Age began in the early nineteenth century, it became increasingly clear that the current material of choice, cast iron, was going to limit progress. Earlier railway work undertaken in the eighteenth century by pioneers like Richard Trevithick foundered with cast-iron rails that were unable to take the loads imposed. Similarly, it could be argued that the early steam road-carriage produced by Trevithick prior to his railway exploits, failed to achieve its potential largely because of the massive components (like the 7ft/3.2m wheels) required from cast iron. Coupled with the appalling road system in England, neglected since Roman times, perhaps steam was abandoned prematurely for road transport. A century later, the internal combustion engine gained a foothold never since relinquished, and has now benefited from over one hundred years of incremental development to reach its present remarkable status.

Despite the limitations of cast iron, especially its lack of ductility in tension, the steam era progressed but not without the catastrophes noted earlier, until the development of economically produced steels as pioneered by Henry Bessemer in the mid-nineteenth century largely saved the situation. This was timely, with the railway revolution really getting underway after the Rainhill trials in 1829, famously won by George Stephenson's *Rocket*, prior to construction of the Liverpool to Manchester line – itself a remarkable engineering feat.

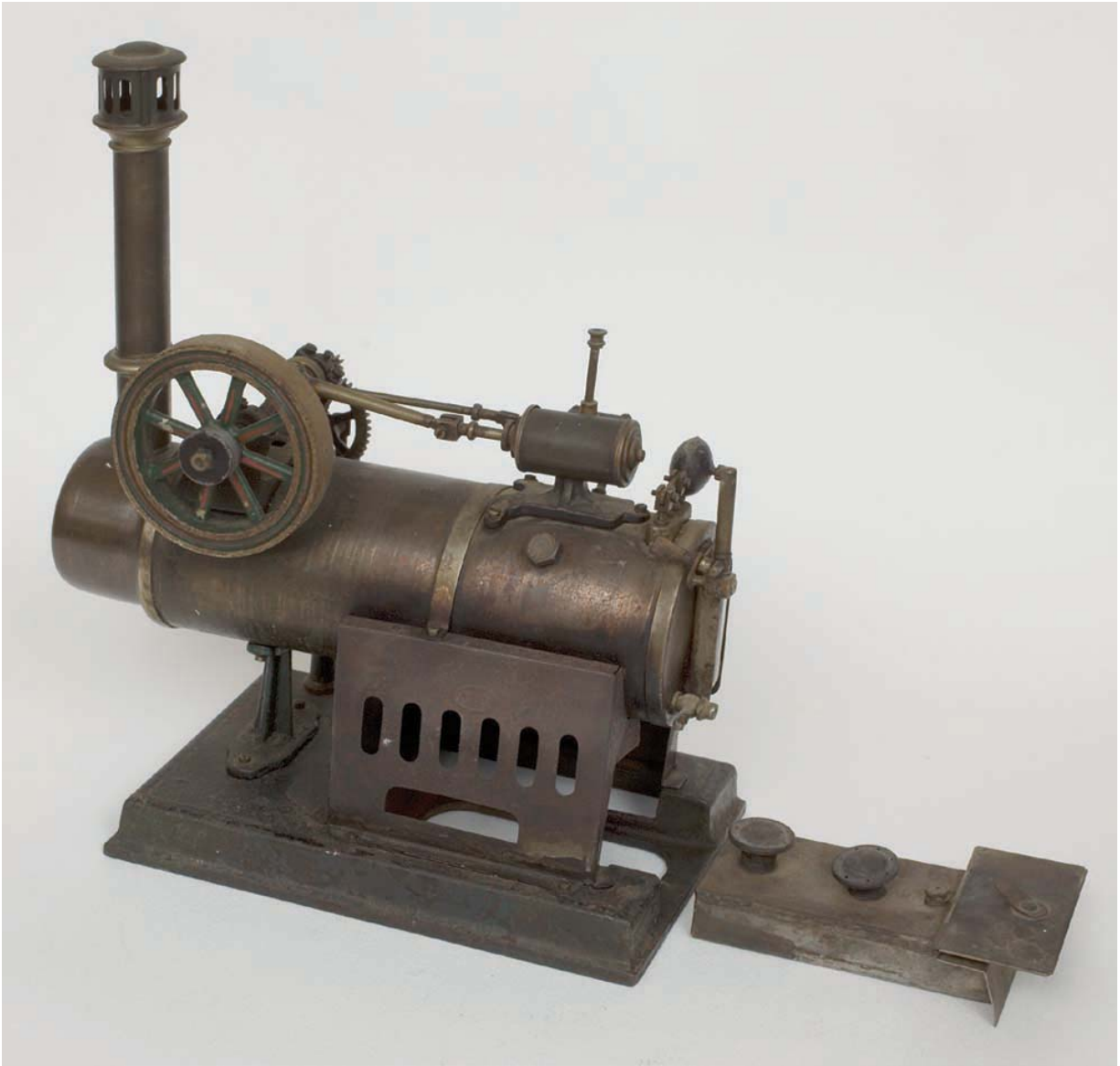


Photo 10 c. 1905 German manufacture, possibly Johann Falke, fitted with injector pump from power take-off, for continuous operation. Often used on trolley to mobilize for farm contract use, such as threshing.



Photo 11 Detail of photo 10, gravity operated pressure relief valve, simple and effective!



Photo 12 Further detail of photo 10, injector pump and its exhaust.

Non-ferrous metals

During the nineteenth century, great advances were being made in the development and production of non-ferrous alloys. Brass and bronze were being produced in tubular form, importantly for the heating elements of steam boilers. Castings could be produced in extremely large proportions as the great Victorian engineers like I.K. Brunel pushed the boundaries of the possible with audacious designs culminating in the gigantic 'Great Eastern'. Almost too large to be launched, it was too far ahead of its time to be wholly successful, but it later enabled another remarkable feat with the laying of the first transatlantic telecommunications cable in 1858 which, despite its composition of 98.5 per cent purity copper, achieved only 50 per cent of possible conductivity – leaving plenty of scope for material improvement!

The latter part of the nineteenth century saw the greater use of aluminium, discovered in 1827, although not deployed as a structural material until the first 'Dural' was produced in the aircraft era of the early twentieth century. Important alloys based on nickel were produced using the Mond process, as devised by Ludwig Mond in 1890 and then used commercially to convert nickel (Ni) oxides into pure Ni by utilizing the unique property of Ni to form a carbonyl compound in the presence of carbon monoxide. A three-stage process enables layers of Ni to be deposited onto Ni pellets. This was once also used for Ni plating work, but is too toxic a process and now 'Electroless Nickel Plating' (ENP) is used. Together with the rarer elements like tungsten, soon to be universally employed in the new electric light filaments, these alloys were produced on a commercial scale for use in the expanding range of alloy steels and high-performance metals. High strength, temperature- and creep-resistant alloys became the materials that enabled successors to the Steam Age to power the industry and transport of our modern world. The fact that over 90 per cent of the present world power is generated through steam turbines, however, arguably indicates that we are merely in a *different* steam age...

Materials testing

Great advances in materials testing were made as these newer alloys were developed – enabling strength, hardness, ductility, toughness, fatigue, creep and corrosion to be measured, even if a rigorous scientific understanding was not generally available until well into the twentieth century.

More recent developments have enabled ever-more sophisticated measurements to be made, but the basic principles have been long-established. The advances in examination of the microstructure of metals, achieved with the advent of the (transmission) electron microscope (TEM), could almost be compared to the unravelling of the mysteries of DNA in biology. Such is the power of resolution on the atomic scale to confirm, for instance, the theory of dislocations

that explains the vast difference between the theoretical and actual strength of metals. The scanning electron microscope (SEM) and its derivatives provide remarkable depth-of-field, enabling a view of fracture faces at multi-thousand times magnification, so that the fine detail of fatigue fracture can be observed. A vital tool in understanding modes of failure, it has been extensively used in accident investigations to determine the difference between mechanical failure and sabotage.

Many processes, like the vital strengthening of aluminium alloys upon which much of the aircraft industry has been based, have been observed and theories postulated – unproven until the minute particles responsible for pinning dislocations during the heating process of ‘precipitation hardening’ were seen with the TEM. Even our old friend, the Fe-C (iron-carbon, as steel) alloy system has yielded remarkable complexity with its variety of phases that can provide a wide range of properties from such apparently simple variants as ‘EN 8’ (ASTM 1040, or 0.4 C steel), providing an easily worked, almost mild-steel strength material, to a tough, a medium strength, or quite high-strength/low-ductility product. We shall see later how this is achieved and how alternative alloys can complement a particular application – the higher alloys primarily being required to achieve properties in larger items, the ‘ruling section’ effect. Thus the secrets of the heat treatment of steels first tackled by the ancient Greeks and Romans have yielded to modern science.

Iron and steel

Iron has been known to man since ancient Egypt, with necklaces of iron beads found in pre-dynastic tombs, where it was clearly rated as more valuable than their precious stones. However, these were obtained from meteorites and easily identified by their high nickel (Ni) content, typically 10 per cent Ni. Early Eskimos had also found similar iron, and this possibly explains the perpetual association with

the ancient gods, like Zeus and Thor, who have been linked throughout the ages with metalworking and blacksmithing.

These early finds were of no engineering importance whatsoever, but by 2,000BC the Egyptians were displaying foundry-work in their tomb paintings. The early bifurcation of metals knowledge eastwards led to the Indian 'Wootz' iron, produced directly from iron ore subjected to laborious hammering and heating in charcoal. This diffused sufficient carbon to enable heat-treatment by quenching, as we would today. Wootz metal, despite little progress for a thousand years, provided steel forgings the envy of the world until the crucible process was perfected in the west by the mid-nineteenth century. China was, again, far ahead with cast iron produced some 1,500 years before the west. Even until the mid-sixteenth century in the west it was normal for there to be no specialization in the manufacture and application of metals – remarkably one man would often prospect the ore, manage the mine, design the smelting furnaces, supervise the foundry and blacksmith the product – and likely even finance the business and engineer the application. Talk about multi-tasking, but little wonder that progress was slow!

Until the Industrial Revolution the 'demand pull' was insufficient to force the pace of iron production, with small-scale production deemed adequate to produce 'wrought iron', analogous to present-day mild steel with ultra-low carbon content, and thus offering no scope for direct hardening. Where strength was essential, as for swords, the time-consuming business of heating in a carboniferous packing was used, similar to present-day 'case hardening'. Indeed, the Greeks were familiar with this process, as alluded to in Homer's *Odyssey*, where the giant Polyphemus is blinded by the grizzly quenching of Odysseus's sword.

By the late fifteenth century in Europe it had become possible, with forced draught, to achieve the temperatures required for iron to be produced in the fully molten state, thus providing cast iron for the first time in the West. However, this was not adopted on an industrial scale until the demands of the Industrial Revolution some two hundred years later.

In 1740 the clockmaker, Huntsman, was perhaps the first to produce molten steel to overcome the unreliability of supplier quality. As Huntsman's steel contained 0.6 to 1.5 per cent carbon, however, it proved too difficult to work and expensive for general use. Not until Bessemer developed his innovative 'Bessemer Converter' process in 1856 did the supply of steel, rather than cast iron, become economically viable, and then not without some traumatic struggles to finally perfect the process. However, the tenacity of the great Victorians like Bessemer, Thomas and Gilchrist, experimenting with furnace linings and different iron charges, finally produced the reliable and cheap steel that we would today recognize as (naturally) – mild steel. The process took the familiar cast iron, of around 2 to 4 per cent carbon, and by forcing oxygen through the molten charge, reduced the carbon to around 0.2 per cent. Thus was provided the strong and ductile material perfectly suited to manufacture of the high-pressure steam boilers, when used in a good engineering design based on the new science of stress analysis, which powered the Industrial Revolution. This material overcame the weakness in tension of nineteenth-century cast iron, but not until much later, with the Liberty Ships in the Second World War, was it appreciated that even these low-carbon steels could suffer devastating brittle-fracture under particular conditions of stress, low temperature, and design, as will be seen in [Chapter 4](#).

After the basic Bessemer Converter process became established and the Victorian entrepreneurs made their fortunes, the range of steels expanded rapidly with the [introduction](#) of the many, often rare, elements to produce the range of 'low alloy' steels familiar today. We will see later how a small palette of these alloys, from the several thousand listed in present-day specifications, is quite sufficient to meet the requirements of even the most demanding task, with the aid of careful heat treatment. Although data is fragmented, sufficient information is provided to achieve the requisite ends, not normally available in texts for the practical worker.

Where corrosion resistance is of concern, and coatings are not the solution, we turn firstly to the high alloy 'stainless steels'. In 1912, at Krupp in Germany, it was discovered that a minimum of 10

per cent chromium (Cr) in steel conferred a remarkable resistance to corrosion, leading to the austenitic stainless steels, such as the '18/8' (Cr Ni) familiar as cutlery. At the Portland Works, Sheffield, the world's first stainless steel cutlery was produced in 1913. Since then, the range of stainless steels has expanded enormously, in order to satisfy the demand for use in a wide range of corrosive environments as, perhaps surprisingly, austenitic stainless can be rendered useless in certain important applications, including some marine uses. Austenitic stainless steels, such as the familiar 304 or 316 types, can be susceptible to crevice and pitting corrosion in the presence of environments such as chlorides, as found in seawater. Another risk is Stress Corrosion Cracking (SCC) in similar environments and where there is a tensile stress, producing a brittle failure mechanism that, ironically, is not normally found in rust-prone steel.

By the twentieth century, steels had become an integral part of the fabric of life, but this is far from meaning that, even including low- and high-alloy steels, they provide the solution to all engineering requirements. Indeed, world consumption is still outstripped by cast iron, itself now existing in many forms, as described in [Chapter 2](#).

1 Case studies: materials in action

This chapter introduces materials used in three practical applications: model locomotive or steam engines ([Case Study 1](#)), a small model aero engine ([Case Study 2](#)), and a two-stroke motorcycle engine ([Case Study 3](#)). It is to be hoped that these items will be of interest to the amateur engineer: often the steam engine from earlier and later life; the aero engine from sore fingers flicking propellers when growing up; even the small two-stroke motorcycle might resonate with those Baby Boomers who were sufficiently fortunate (or otherwise) to encounter them as early personal transport, even if this particular model's failings and missed potential reflect on the demise of this sector of UK manufacturing.



Photo 13 Dave Waggett of the Stockport Model Engineering Society powers around their track.

CASE STUDY 1: MODEL STEAM LOCO BOILER

Materials

The copper loco boiler is the heart of the model steam train. A similar construction can be found in model traction engines, and simplified examples are employed widely in many and various other steam plant, examples of which are illustrated in the [introduction](#) to this book.

The fabrication of the loco boiler is quite a demanding task, not to be undertaken before sufficient experience has been gained in the fields of rolling and bending plates, and the requisite joining methods. Design is also paramount; many established and proven

designs are readily available. The importance of structural integrity must not be underestimated as mandatory requirements exist for the pressure testing and operational safety, as high-pressure gas (steam) provides a potentially dangerous energy source. Indeed, the catastrophic record of full-size boiler failures throughout the Industrial Revolution provided the drive to understand the design and manufacture of pressure vessels, which we explore in [Chapter 3](#).

Therefore, boiler fabrication should be treated with due respect, but with proper experience it does provide a perfectly practical activity for the home worker, and there is no shortage of information with which to tackle the subject (see [Bibliography](#)).



Photo 14 Boiler build nearing completion in Ron Perry's workshop.

Table 2 Copper for boilers.

Grade		N/mm ² UTS	0.2% PS	(%) Elong ⁿ	Hv	(oC) MP
BS	C106	200-400	50-340	50-5	40-120	1083
UNS	C12200	} similar properties				
CEN	CW024A					
ISO	Cu-DHP					
ASTM	C12200					

We are primarily concerned with the materials of construction here, and it is worth noting the approach taken by makers of model loco boilers in the almost universal choice of copper (Cu) for the main construction. This contrasts with the full-size machines, which almost universally employ plain carbon steels for boilers, generally relying on riveting for joints, as arc welding was not available for practical application until early in the twentieth century.

Whilst there exist alternatives for model boilers, such as plain carbon steels and stainless steels, the consensus of makers indicates that copper remains the optimum choice. This is primarily driven by the exemplary corrosion resistance, important with the thicknesses involved, typically 3mm ($\frac{1}{8}$ in), whereas the full-sized version uses relatively thick plate upon which a corrosion allowance may be applied without significant economic or practical impact.

Stainless steel might be thought a practical alternative, and indeed it has been tried on occasion, but it presents a more complex choice. For one-off fabrication the selection of a particular grade of stainless presents significant practical difficulties which would not be the case for high-volume production with optimum material selection. As will be explored later, stainless steels are not necessarily free of corrosion problems, such as pitting, crevice and stress corrosion cracking in particular environments, which could be the case with certain boiler operations.

The grade of copper most commonly employed is generally recognized as 'C106', which relates to a superseded British

Standard but, as explored later, is still common currency in practice. The equivalents include UNS C12200 and, in the USA, C12200, such as ASTM B75 for seamless tubing. This is a 'phosphorus deoxidized non-arsenical copper', thus avoiding potential problems with oxygen embrittlement, with properties shown in [Table 2](#).



Photo 15 Firebox view of Photo 14, showing reinforcement universally employed.



Photo 16 LMS Duchess, showing varieties of silver soldered joints.

Joining

The method employed for fabrication of these Cu bodies is normally silver soldering. At one time it was thought that an extensive range of silver solders, with a wide range of discrete melting points (T_m), were required as the sequence of joints was made, with highest T_m followed by progressively lower, to finish at the lowest without re-melting previous joints. With experience, however, two alloys are sufficient to effect a full and complex boiler construction, as shown in the photos.

For these examples the grades SB40 and SB55 provided the melting points as shown in [Table 3](#). These alloys are now cadmium

(Cd) free, comprising essentially Cu-40 per cent silver (Ag), and Cu-55 per cent Ag. This provides around 50°C difference in T_m . Careful planning of the job is required to ensure that subsequent joints are far enough away to avoid disturbance of those previously made, and also with sufficient access.

Heat sources usually employ gas-air, such as when using specialist propane torches. It is possible to use oxyacetylene but the concentrated heat source requires experience to avoid overheating, although it can solve some problems of difficult access, and flame extinction in constricted areas.

Careful preparation, with regard to joint fit-up and especially cleanliness, is vital in the making of successful joints by this method, as is sufficient practice. Flux is required, an example being the general purpose 'Easy-flo' powder or paste flux marketed by Johnson Matthey plc (see [Appendix II: Resources](#)). Cleaning is by acid dipping, such as water plus 10 per cent sulphuric acid for about two hours, followed by washing off thoroughly, and drying.-Scrupulously clean joints are essential, and touching by bare hands must be avoided!

A cautionary note should be made in two instances. Firstly, it may be tempting to try a flux-coated copper-phosphorous brazing rod, but this must be avoided as the deposit is susceptible to cracking at the elevated temperatures experienced in the firebox. Secondly, silicon bronze rods have been tried for brazing copper boilers, but in operation this produces an electrolytic action which makes the joint prone to de-zincification and failure (see the 'Galvanic Series' in [Chapter 4](#)). This may proceed at a slow rate, even taking years, but it is now barred from use because of this experience.

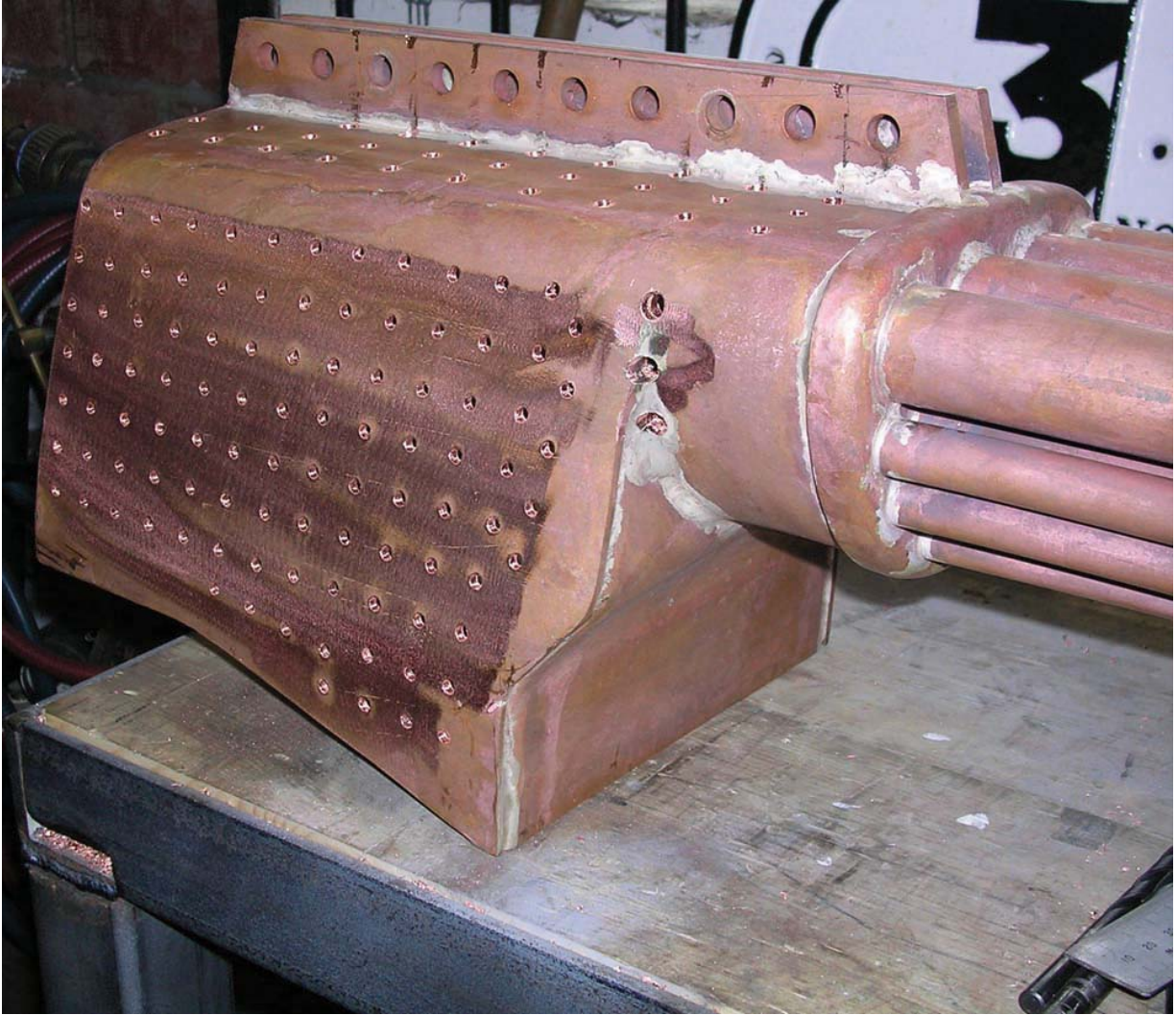


Photo 17 Firebox and boiler tube construction arrangement.

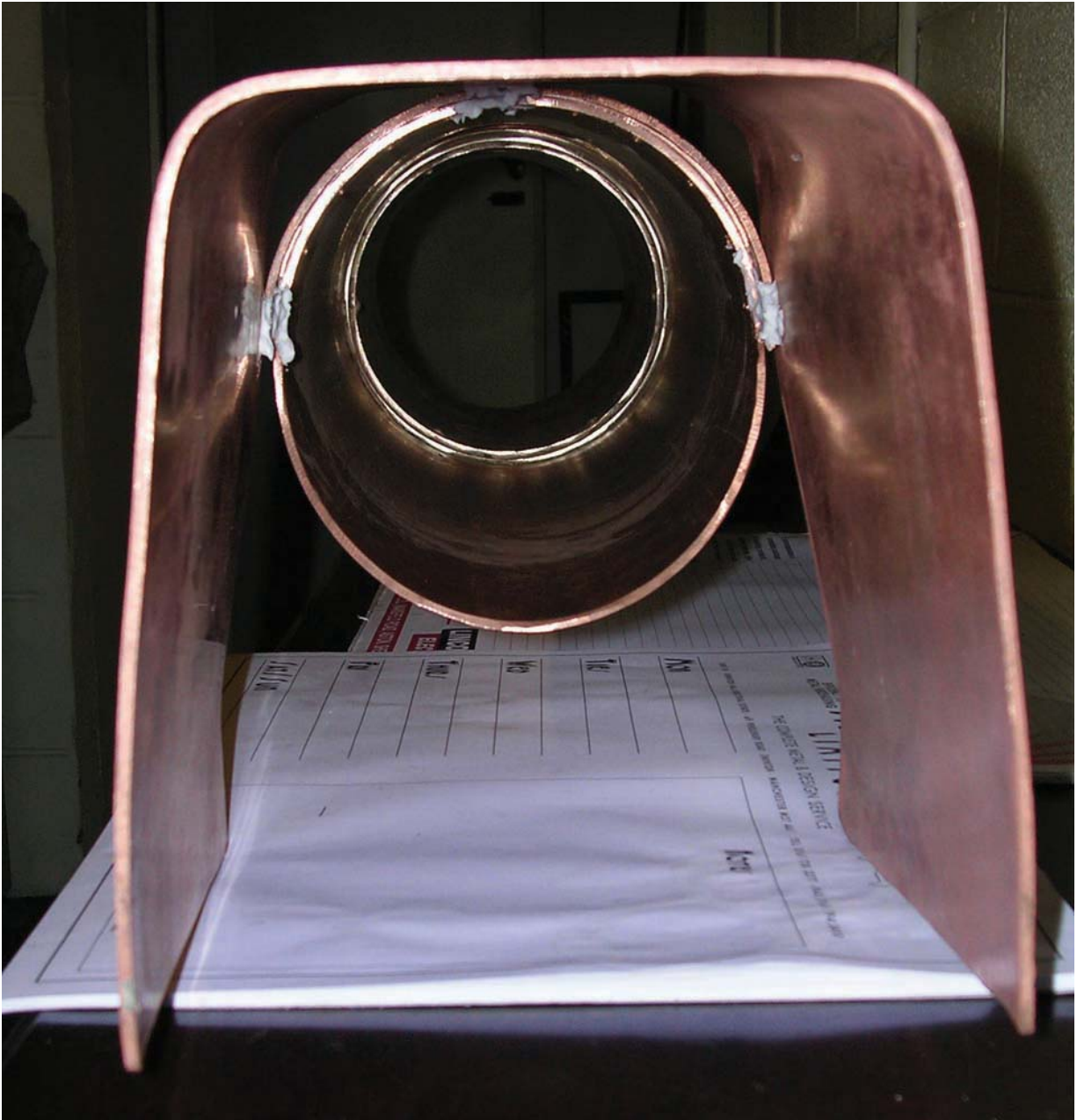


Photo 18 Early stage of fabrication of boiler to firebox main parts.

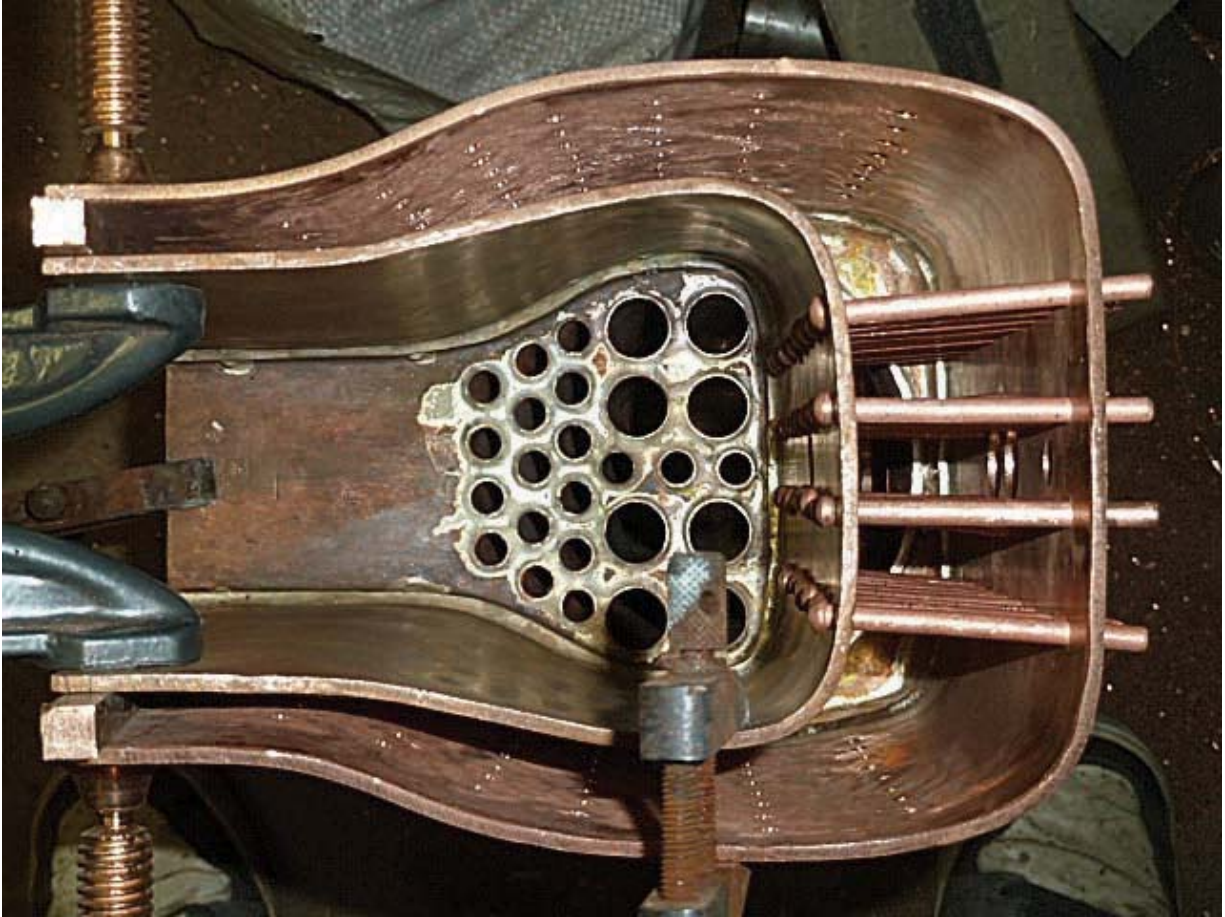


Photo 19 Internal view during fabrication of firebox internals and reinforcement.

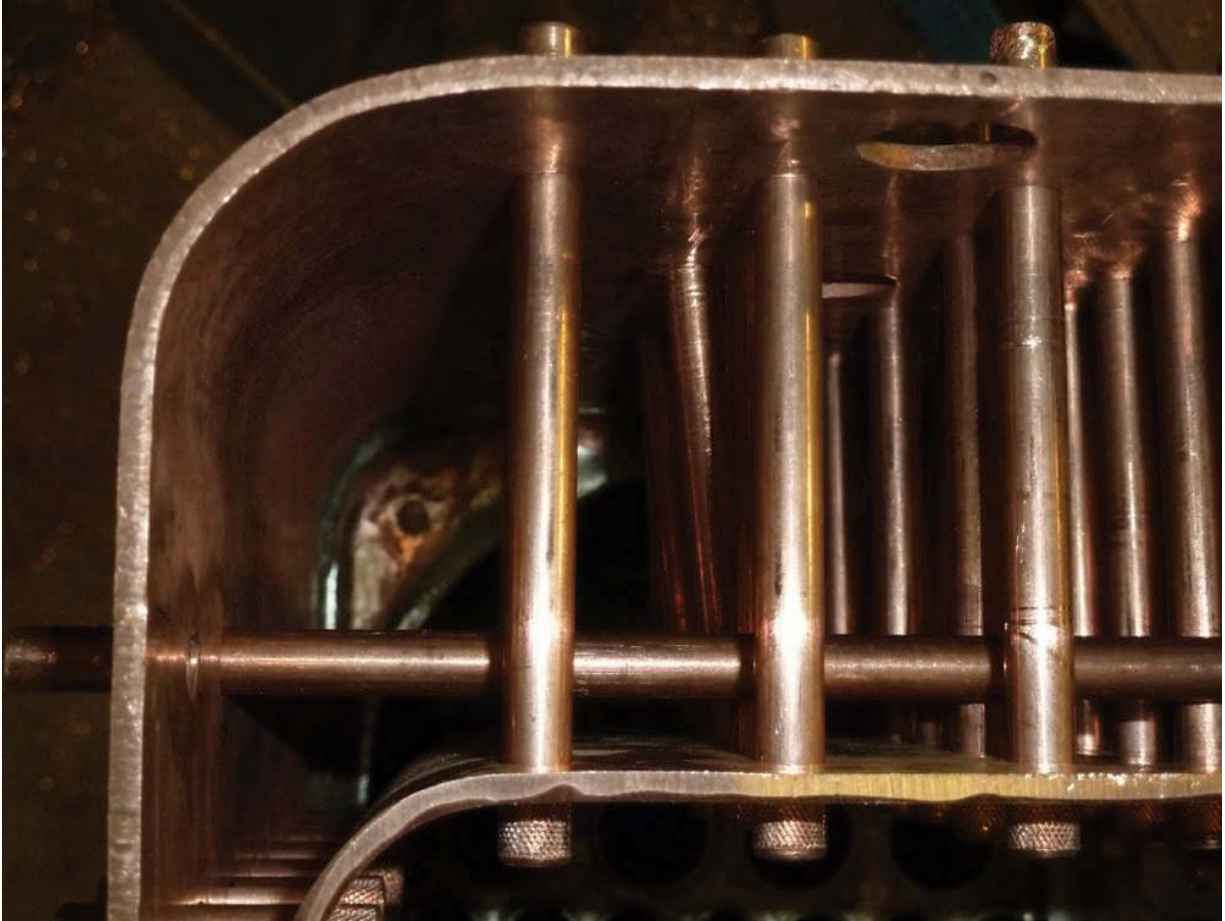


Photo 20 Reinforcement of firebox by stays, prior to silver soldering.



Photo 21 Detail of tube to tubesheet joints, silver soldering complete, boiler side.

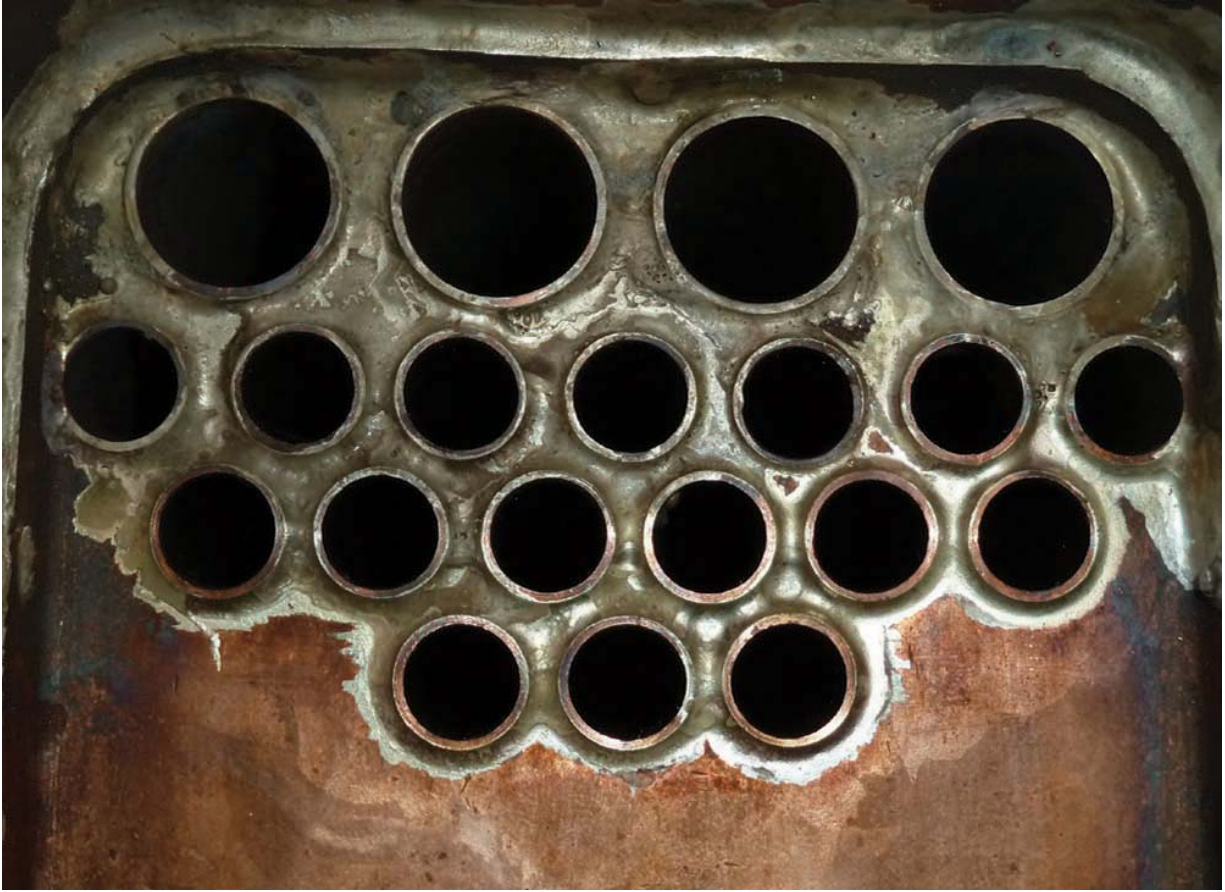


Photo 22 Reverse side of tubesheet, showing penetration of silver soldered tube joints.



Photo 23 LMS Duchess on boiler proof testing, before build. -

Table 3 Silver soldering alloys – tensile (UTS) and shear strength.

Grade	BS1845: 1984	Composition	UTS/shear (N/mm ²)	T _m (°C)
SB40	(Ag14)	Cu-40Ag - 28Zn - 25n	~430/155	650-710
SB55	(Ag20)	Cu-55Ag - 22Zn - 25m	~400/145	630-660

CASE STUDY 2: MODEL AERO ENGINE – 2.5cc

Photo 24 below shows a well used ‘PAW 249’ model aero diesel (‘compression ignition’) of 2.5cc capacity, manufactured by the Progress Aero Works of Macclesfield, Cheshire, UK. This is a fine example of precision engineering, delightfully simple in design and

successfully manufactured by a small independent business, into an exacting commercial market.

Founded in 1948, over three generations its proprietors have held numerous UK and even European championships in the field of speed control-line flying with engines basically similar to that illustrated. Despite its simplicity, the rated power of 0.33 bhp at 16,000 rev/min, weighing 170g, equates to ~133 bhp per litre. The higher performance, twin ball-bearing model develops 0.58 bhp, equating to 233 bhp per litre, of a similar order to that of a F1 racing car! This particularly illustrates that very small internal combustion engines can be remarkably power- and weight-efficient, derived from the extremely high compression ratios and speeds attainable with such a low reciprocating mass. This is, of course, rather unfair on the larger engines required for vehicles that have to operate under different speed ranges, and a 'forty-cylinder'(!) version would hardly be a practical alternative for the 35 bhp per litre engine of [Case Study 3](#).

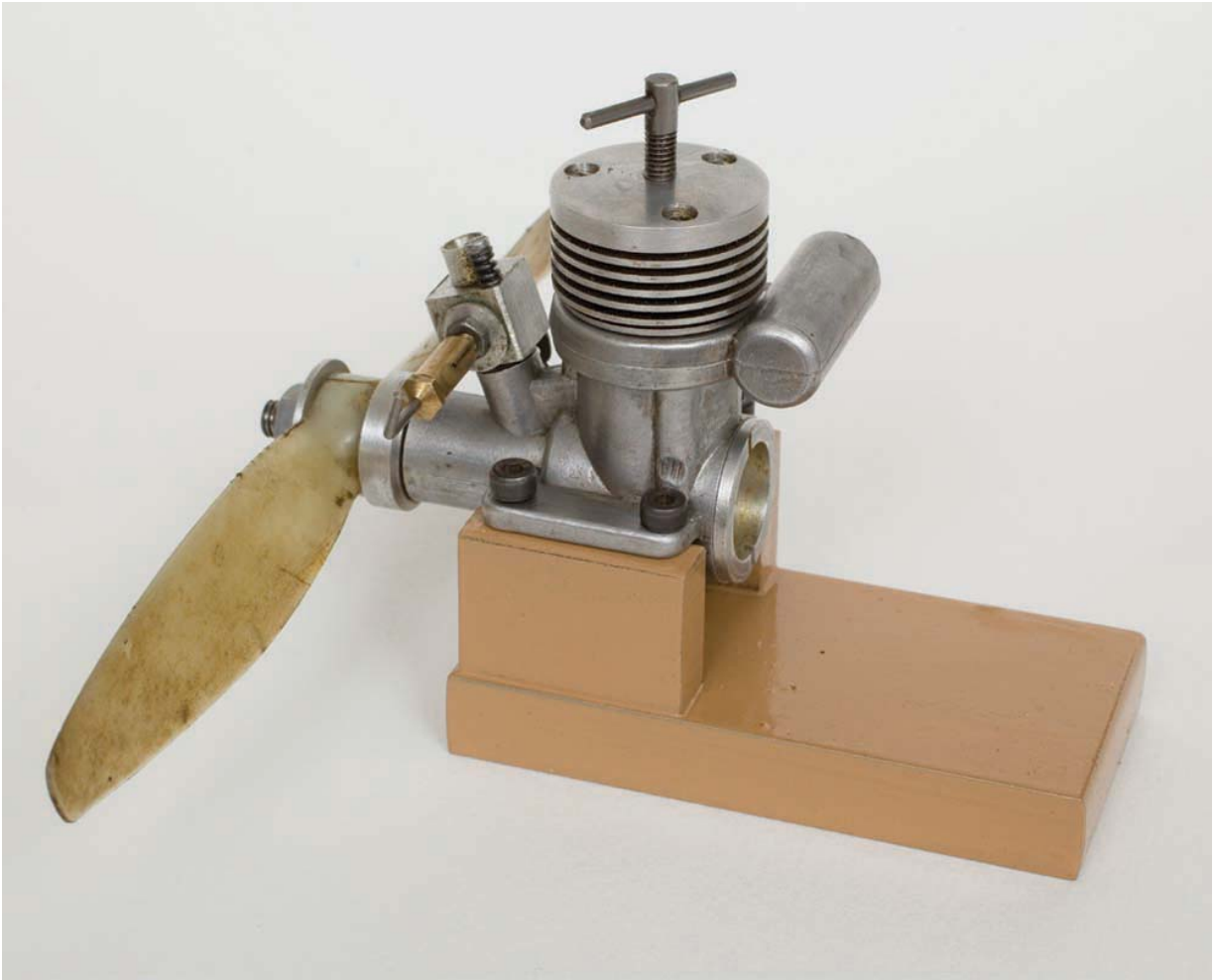


Photo 24 PAW 249 model aero engine in test jig.

The intrinsic simplicity of design possible with this type of engine has long made it an attractive prospect for the amateur engineer with a bent for very high precision, small scale work. For a commercial model however, selection of materials and design requires great care to achieve the required performance without excessive assembly time.

[Photo 25](#) shows the essential elements of crankcase, cylinder barrel, crankshaft, piston and con-rod, fuel delivery unit, and fuel metering needle valve. A major feature is the weight of each component: the static parts, as in crankcase and cylinder, where their contribution to overall power-to-weight ratio is key; the moving parts, such as crankshaft and piston; and the con-rod, where inertial loading at these ultra-high speeds becomes a material-limiting factor.

As this is a commercial, relatively low-cost engine, the material selection must also be sensitive to price, availability and machinability – as well as the mechanical properties of strength, toughness and stiffness, as discussed in [Chapter 3](#).

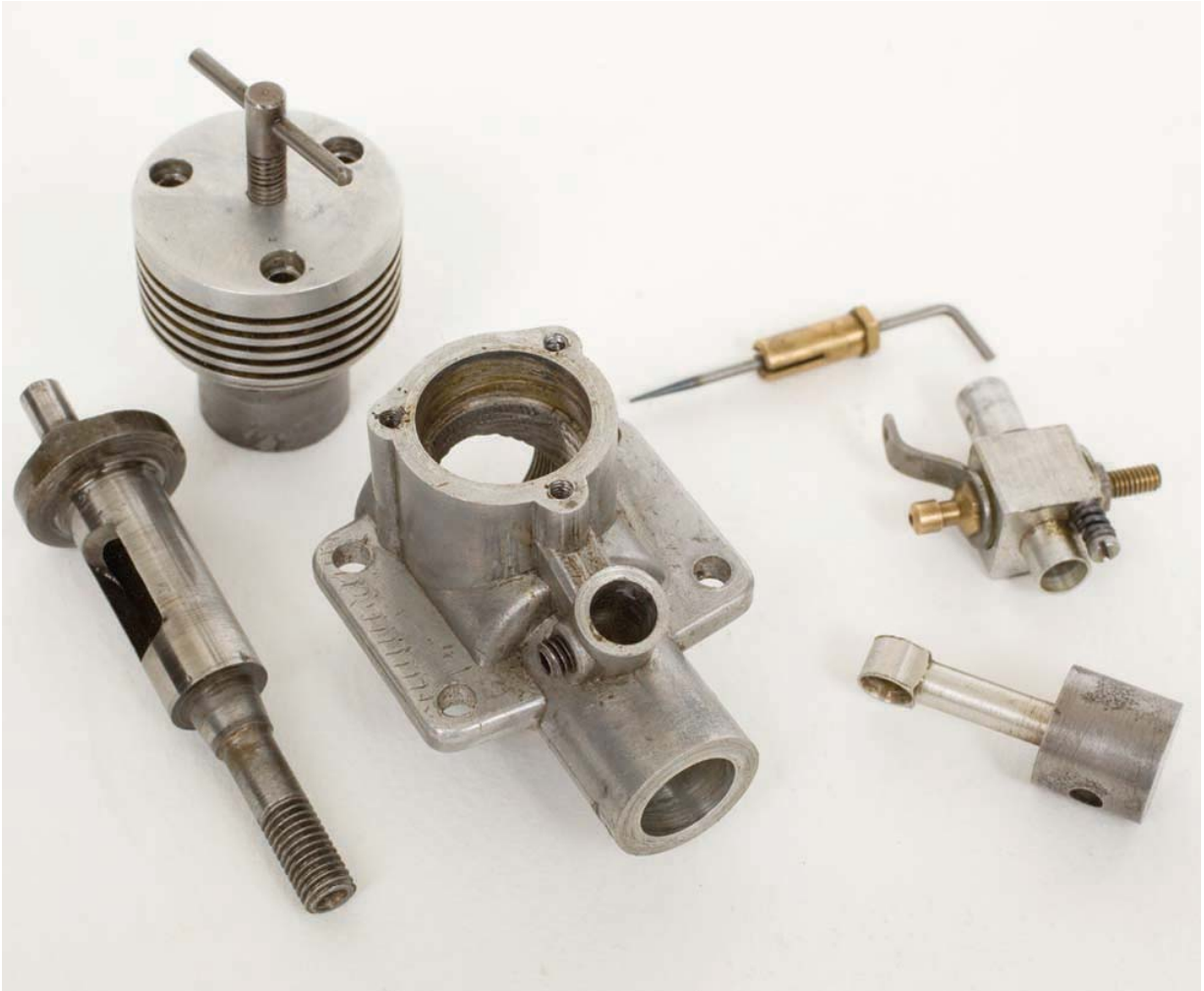


Photo 25 Main components of PAW engine, as examined.

The crankcase (see [Photo 26](#)) is produced in-house by casting the aluminium alloy into a metal die. Unusually, this die was also designed and manufactured in-house, providing a high quality, precision casting, requiring only internal machining. The melt is provided from a 50:50 preparation of charge material from LM4 and LM6 alloy (see [Table 4](#)). The LM4 alloy is aluminium (Al) containing 2 to 4 per cent copper (Cu) and 4 to 6 per cent silicon (Si), which

presents the possibility of age-hardening (see [Chapter 5](#), Back to Basics, for the theory of such strengthening). This strength bonus is not utilized in production as the as-cast strength is sufficient, which in theory provides potential for a small weight saving by section reduction. The LM6 is a familiar Al-12 per cent Si casting alloy, often found in intricate castings requiring high fluidity. The combination is clearly successful in meeting the requirements of castability without the need to achieve the highest strengths required in primarily structural components.



Photo 26 Crankcase, onepiece aluminium alloy die casting.

Table 4 Aluminium casting alloys, LM4 and LM6.

Element	LM4(%)	LM6(%)

Al	87-91.4	85-88
Cu	2.0-4.0	0.1
Si	4.0-6.0	10-13
Fe	0.8	0.6
Pb	0.1	0.1
Mg	0.2	0.1
Mn	0.2-0.6	0.5
Ni	0.3	0.1
Sn	0.05	0.05
Ti	0.2	0.2
UTS (N/mm ²)	>160	>190
Elongation	2%	7%
Density (g/cm ³)	2.77	2.68

The cylinder barrel ([Photo 27](#)) comprises an aluminium alloy finned section, pressed over a steel liner, the latter forming the transfer and exhaust porting whilst accommodating the piston. The finning is generated from a turned and bored bar of 2011 heat-treated alloy, providing strength, although the material is chosen primarily for high heat transfer and low density, with good machinability as an incidental benefit.

The cylinder 'liner' is machined from EN8, quench and tempered steel as detailed in [Chapter 2](#). The 'transfer' ports are machined direct into the bore, and the piston effectively forms the inner wall. Whilst on large engines such a liner is often chosen from a grey cast-iron, for such a small unit the toughness, strength and wear properties make the steel a practical choice, combined with a requirement for operation at elevated temperatures.

Steel EN8, quench and tempered, is also chosen for the heart of the engine and the crankshaft which drives the propeller; it also forms the inlet gas valve, and crankpin onto which the con-rod locates, as shown in [Photo 28](#). The principal requirement is strength and toughness, with good machinability from the bar-stock supplied. The central portion provides the bearing area, with later models using the higher performance twin ball bearings instead of this long plain bearing. This area is cylindrically ground to provide the fine surface finish that provides the low-wear characteristics, running in the softer housing of the crankcase, as discussed in [Chapter 4](#). The effective mating of these surfaces is also critical in effecting a seal, as the crankshaft aperture provides the inlet port during each cycle. These surfaces are treated to a commercial 'carbo-nitriding' process to optimize surface hardening and reduce wear.



Photo 27 Steel cylinder liner and finned aluminium alloy jacket, transfer ports as grooves in liner, exhaust ports clearly seen.



Photo 28 Steel crankshaft, plain bearing type, showing large inlet port.

A highly-loaded rotating component, such as the crankpin, could be subject to early failure by fatigue were not the material selection and design sound, as discussed in [Chapter 4](#). This part is subject to a complex cyclic regime of bending and torsional stress. Strength, combined with lack of crack sensitivity, is required for a design with a safe fatigue threshold and long design life.

[Photo 29](#) shows the piston, gudgeon-pin, and con-rod assembly. In this model the piston is a grade 17 grey cast iron, as described in [Chapter 2](#), selected for low sliding-wear in the cylinder. But it has a relatively high inertia and for higher performance, use of an aluminium alloy piston is required. The small size makes use of compression rings, universal in larger engines, impractical for commercial units. However, the material combination used has a

proven record of sealing effectiveness without the need for piston rings.

The gudgeon pin is a simple pin through the piston. This was previously made from a quench and tempered 'silver steel' (see [Chapter 2](#)), but now employs a standard steel dowel pin, so avoiding the difficulty in precise machining of such a small item.

The con-rod is machined from a 2011 heat-treated bar of sturdy dimensions, to withstand the high compressive loading imposed by the high combustion pressures of a compression-ignition (diesel) engine, whilst limiting the inertial loads due to the high rotational speeds.



Photo 29 Cast iron piston and robust aluminium connecting rod appropriate for high compression ratios.

Finally, the fuel inlet and metering valve, shown in [Photos 30 and 31](#), comprise an aluminium (2011) alloy body machined from square section, with a thin aluminium alloy tube forming the venturi, across which is inserted the proprietary brass valve which takes the fuel to the venturi jet. A simple rotating valve is employed to control in-flight fuel supply, and thus motor speed.

This valve mechanism is made from a 60/40 (Cu-Zn) brass, as is the threaded 'nut' part of the needle-valve, the centre part of which is a taper-ground rod of silver steel, with cranked end to facilitate fine adjustment of the fuel supply, for tuning the motor for optimum performance.

A remarkable small engine, of proven design and durability, from accessible materials – reinforcing the old adage, 'what works is right'.



Photo 30 Spray unit, analogous to a carburettor, for fuel supply to inlet port, with controls for speed variation and cut-off.



Photo 31 Needle valve used to meter fuel supply.

CASE STUDY 3: CLASSIC TWO-STROKE MOTORCYCLE ENGINE – 150CC

This engine is the basis of possibly the most popular UK motor of its ilk, being produced from 1948 until its departure from the once-great BSA motorcycles in 1971, with well over a quarter of million units manufactured by the 1960s.

Whilst this 150cc unit, first produced in 1954, is rated at a lowly 4.9 bhp at 4750 revs per minute – equivalent to ~35 bhp per litre – it was sufficient to drive the machine adequately, and subsequent development showed that it could be improved to over 100 bhp per litre for competition use, and it retains an active and successful following to the present day.

[Photo 32](#) shows the complete unit, innovative at the time for combining into a compact and light single casting the power unit and gearbox, albeit supplying only three gear ratios. Its essential robustness has been demonstrated by subsequent power increases within the same design envelope, and use of the same materials. Despite the low build-cost and relatively high production volumes,

the material choices are sound and provide a safe basis for internal modification, with only minor external changes over its entire production life.



Photo 32 148cc two stroke BSA classic motorcycle engine, on handling stand.

The cylinder head, [Photos 33](#) and [34](#), provides a good example of an intricate sand casting, using an Al - 12 per cent Si alloy such as LM6, where the principal criterion is replication of the heavily finned design which provides the essential heat dissipation required for the

two-stroke, air-cooled engine. The final design of this part copied the work of individual Bantam tuners, with even more heavily finned designs and central plug for optimum combustion, higher compression ratios, and power increase. In fact, the 'man in the shed' had a significant influence on the later engines, with the BSA corporation following individual innovations, a theme more often seen in product development than might be expected, for example the origins of the current BMW Mini Cooper and derivatives.

[Photo 34](#) shows the hemispherical combustion chamber, readily machined with its cylinder-gasket sealing face. The residue of the casting box split shows how the casting pattern has been trapped with the gasket face at right angles to the casting box, providing the most efficient 'runner', 'riser' and 'feeder' for this simple gravity-cast arrangement. There is almost no trace of this centre line on the finned side, which will have been lightly dressed on removal from the mould, to enhance cosmetic appearance (see [Table 5](#)).



Photo 33 Cylinder head casting, showing deep finning possible for economic manufacture by sand casting.



Photo 34 Cylinder head, showing combustion chamber and casting mould separation line.

Table 5 Piston alloys, 4032–T651 and 6082–T651.

	4032	6082
(Element)	(%)	(%)
Al	85	95
Cu	0.5–1	<0.1
Si	11–13.5	0.7–1.3
Fe	<1.0	<0.5
Mg	0.8–1.3	0.6–1.2
Mn	–	0.4–1.0

Ni	0.5–1.3	–
Ti	–	<0.1
UTS (N/mm ²)	360	340
0.2% PS	330	310
Elongation	4%	11%
Hardness	136 Hv	100 Hv

[Photo 35](#) shows the grey cast-iron, medium strength, casting that forms the cylinder barrel. The material choice is a compromise between castability of the complex shape, with optimum wear characteristics, and weight. The weight penalty was clearly considered insignificant compared to the cost of a nearer to optimum design with a cast-iron liner set inside an aluminium finned body, in a similar manner to the PAW engine. In recent times, individuals have successfully produced such designs, but at considerable economic penalty.

The complexity of the design required for the two-stroke engine is clearly seen in [Photo 36](#), where the gas transfer passages are made integral within the casting body, taking the gases from the crankcase to mid-way up the cylinder, with release controlled by piston position, as for the inlet port and exhaust port, by the bottom and top of the piston. All of these passages are provided in the ‘as cast’ condition by the [introduction](#) of casting ‘cores’, as described in [Chapter 2](#). Success in such casting accuracy means that no further expensive ‘fettling’ is necessary. Conversely, as these passages have a major impact on performance, they present the sufficiently knowledgeable individual worker with an opportunity for tuning, at the cost of time-consuming hand grinding and filing. Fortunately, this is aided by the good machinability of the grey cast iron.

A feature of this particular piece is the obvious broken fin – probably resulting from an ill-chosen levering-off of the cylinder head. It does demonstrate the complete lack of ductility of this

material, with the remaining fin showing no permanent deformation around the break. It has simply 'snapped off like a carrot', shortly after the elastic limit was reached, with almost zero plastic deformation, unlike common ductile metals like mild steel, as explored in [Chapter 3](#).



Photo 35 Cast iron cylinder, deep finning and fin fracture with no local plastic deformation.



Photo 36 Cylinder, showing large transfer and exhaust ports, produced by coring and requiring only minimal fettling.

The crankshaft/flywheel/con-rod assembly is shown in [Photo 37](#). Its high inertia contrasts with the PAW engine, the large fly-wheels being augmented by a further flywheel forming the (unseen) magneto, which locates on the shaft via the large keyway. The flywheel inertia of the Bantam is utilized to assist slow running from ~500 to 5000 rev/min as speed is varied and with only three gears to cope with road conditions. The aero engine, by contrast, is designed to operate at or near peak power, at three times the Bantam's

maximum, as is common for full-scale aircraft, without the need for a wide spread of power versus speed capability.



Photo 37 Flywheel/crankshaft/connecting rod assembly, showing standard mild steel flywheel discs to assist performance.

Despite this limited original operational range, the same basic design has been shown capable of handling twice the original rotational speed, with removal of the flywheel-magneto. The main flywheels are produced from a cast steel as two large discs, with the original casting side-profile hidden by the mild steel discs. These are retained in a small recess near the flywheel rim, by 'pop marks' to

prevent them from coming loose in action. Ironically, in some of the later models, BSA re-applied similar discs as part of a performance-enhancing programme. Unfortunately, they were retained by rivets which habitually came loose, the inertia throwing them out like bullets from a machine gun, with predictably disastrous damage to both engine and maker's reputation (a fine example of corporate loss of previous experience...).

The shafts are fine machined, by cylindrical grinding, from low alloy steel, such as EN24, heat treated to near maximum strength. These are seriously stressed parts, but have proven free from fatigue failure despite the severe stress concentrations, such as the large keyway, proving a good material and manufacturing choice of a critical part. This is, perhaps, an example of something working well where with an over-theoretical approach one might not have dared to go (in other words, 'paralysis by analysis'). For sure, this part will not have been subjected to the kind of sophisticated stress analysis available today, which might have produced a more (unnecessarily) robust design.

The con-rod is an assembly of 'big end' bearing and crankpin, basically cast steel for the former and low alloy steel (EN24, again) for the latter, with a roller bearing, all pressed into the flywheels under a heavy interference fit. This effectively puts the mating parts under high local residual stress, to ensure no movement of crankpin to flywheel under action. (See [Chapter 3](#) for residual stress discussion.)

The con-rod 'small-end' shows a relatively soft bronze bush pressed into the casting, providing a hard-working bearing to support the piston, via the hard steel gudgeon pin. The latter is made from a fully hard low alloy steel, possibly AISI 52100 type, a very finely finished, cylindrically ground proprietary item. This small-end bearing is subject to high loads and poor lubrication from oil entrained in the fuel, and is subject to wear. In later engines a needle roller bearing is used, as a sensible improvement.

The crankcases shown in [Photos 38](#) and [39](#) comprise, essentially, a sand-cast aluminium alloy based on LM6, with housings for the

main bearings to accept the crankshaft, plus the integral gearbox with its complex arrangement of gears and selector mechanism.

The aluminium castings show a good example of the pattern maker's and foundryman's craft, with good replication of detail and bearing faces, before machining. These have proven very robust units, even when subjected to considerable power improvements. Later versions employed die-castings for improved productivity and even better surface finish, but were not as robust as the original sand castings. The material choice is sound, providing sufficient strength, ability to retain accurate machining without distortion, good heat transfer and low density, all coupled with cost-effective production.

The bearings seen in [Photo 39](#) are an interference fit in the casting, use being made of the high thermal expansion of the aluminium, enabling heating above operating temperature to provide a tight fit on assembly, and at operating temperature.



Photo 38 Aluminium crankcase half, with interference fit ball bearings.



Photo 39 Crankcase internal view, showing casting detail.

[Photo 40](#) shows the final-drive gear which forms part of the gear train, all of which is manufactured from a medium heat-treated low alloy steel like EN24, requiring high strength *and* toughness to withstand the normal gear tooth stresses, plus the gross overloading imposed during normal use and abuse of these critical parts. Wear is dictated by gear engagement design, material selection and operation conditions (lack – or otherwise – of abuse!). These items show no significant wear, a testament to the design and material selection. For present-day use, such engines really require different ratios and additional gears – perhaps the most testing project for the home worker, not lightly undertaken.

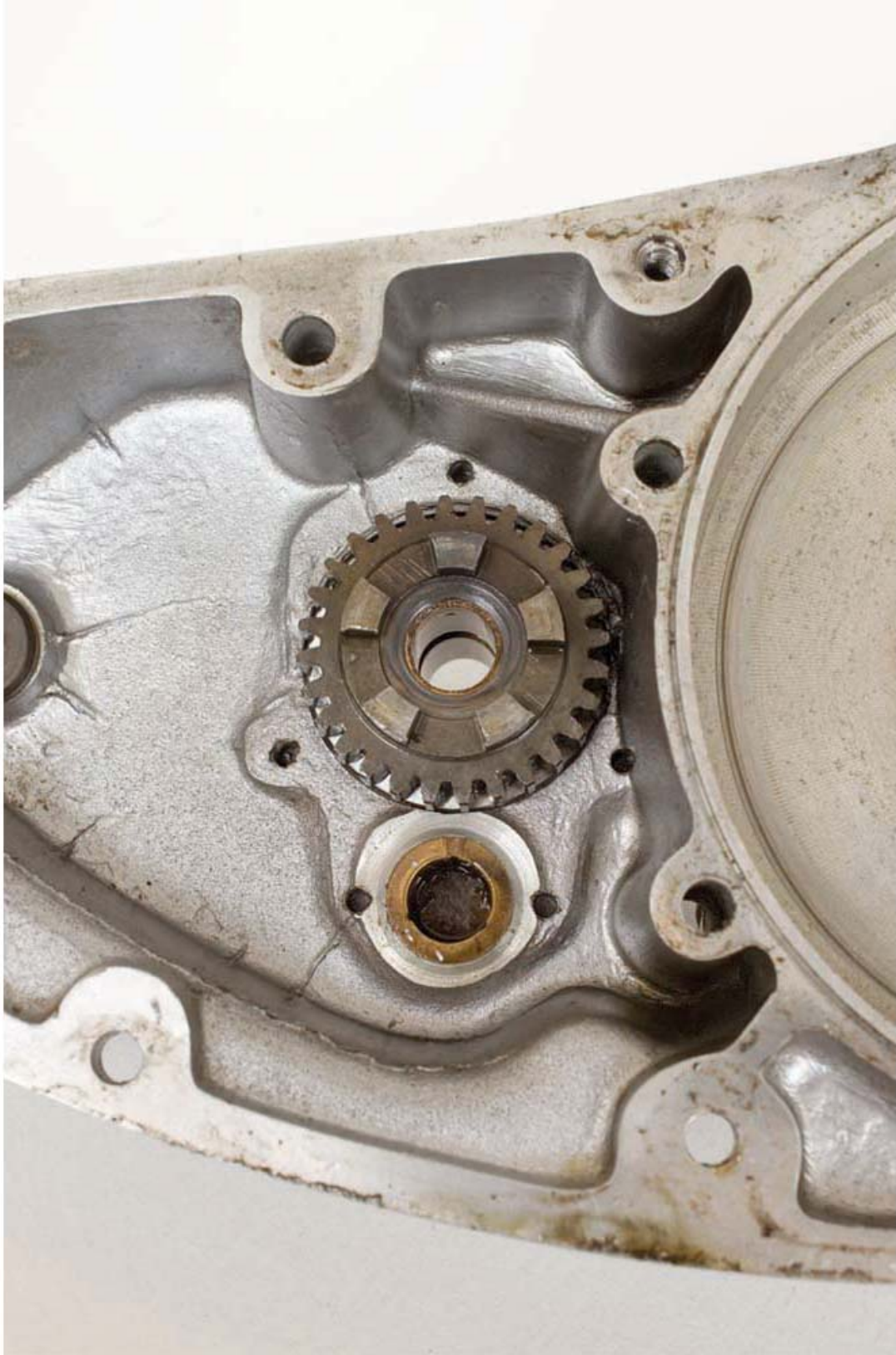


Photo 40 Integral gearbox, internal view showing low alloy steel top gear pinion with lack of wear on engagement dogs demonstrating good balance of hardness and toughness.

Photos 41 and 42 also show the gear-selector mechanism, its complex shape provided most economically by a special steel casting, although this is not as highly stressed as the individual gears and shafts, all of which demand very careful control of heat treatment and machining in order to provide the precision essential for efficient running and loading, with full interchangeability.

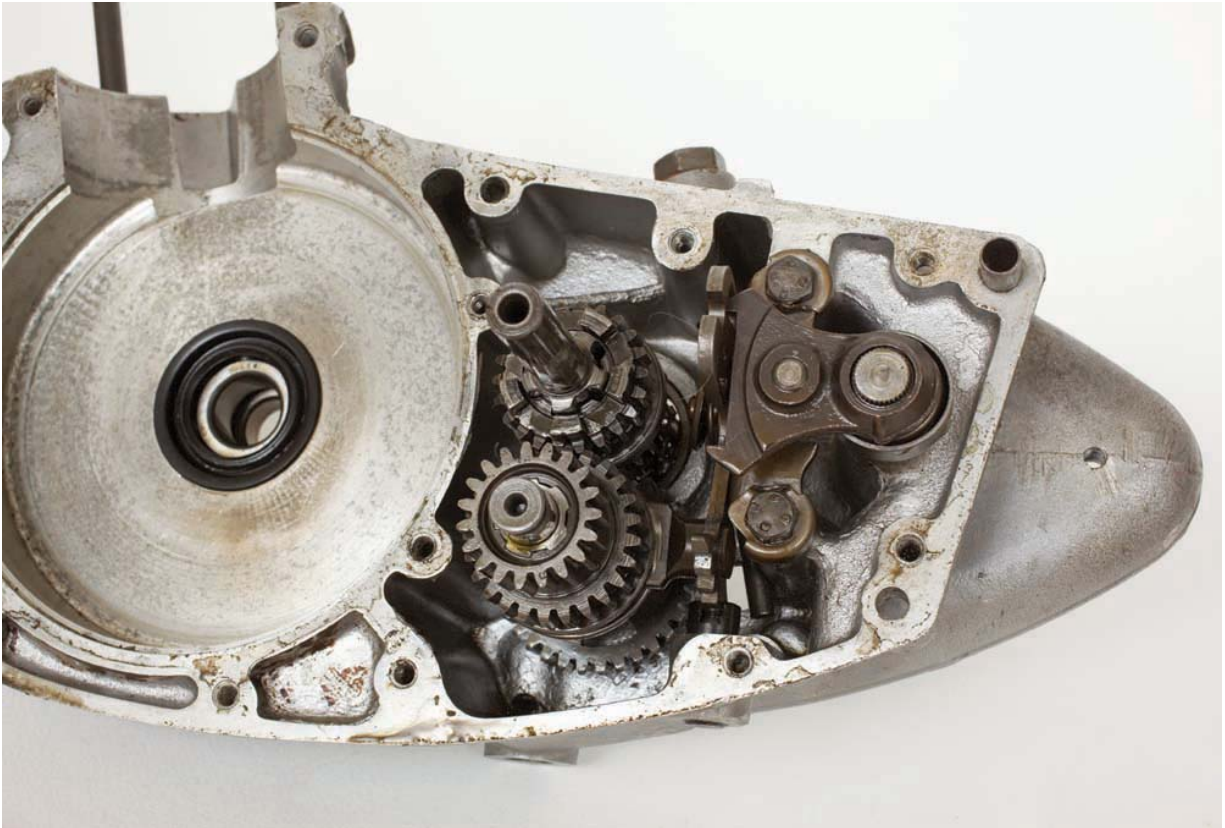


Photo 41 Crankcase half containing gearbox assembly.





Photo 42 Gear pinions and shafts, using low alloy steels.

The piston shown in [Photo 43](#) presents an opportunity for several aluminium alloy choices, with the aluminium base being the universal choice because of its low density and high strength-to-weight ratio. The detail choice is basically an Al-12 per cent Si near-eutectic alloy suitable for casting, with a relatively low thermal expansion coefficient, and sufficient toughness. The expansion ratio, especially in a two-stroke engine with its marginal lubrication of piston/cylinder surfaces, is important in order to maximize the pressure seal, and reduce 'knocking', whilst avoiding 'seizure', where the clearance becomes insufficient to prevent a temporary surface adhesion. This is clearly seen in [Photo 43](#) as a vertical score mark, potentially damaging to the proper running of the engine. Severe cases can produce local solid-state fusion, and result in the engine suffering a catastrophic arrest, discussed in [Chapter 4](#) under 'Wear'.

Low Si pistons are typically of the 6082 type, with low Si but magnesium (Mg) and manganese (Mn) addition, which provides high strength, especially in the heat-treated condition. These are similar to the materials developed as the HE30 grade, and 'Hiduminium' RR50 by High Duty Alloys for aircraft engine development in the UK in the early 1950s. (This is a good example of how reference to the USA/International 1000–8000 series designations for the aluminums is far more satisfactory.) Current practice would indicate that a 4032 or 319, Al-Si-Cu alloy, would be the best choice for an enhanced performance piston of this type (see [Table 5](#)).

The piston rings, in contrast, are produced from a low ductility cast iron, in order to provide the sliding characteristics required whilst coping with the lack of plasticity (with the relatively large elastic deflection) by virtue of its small-section strain. These rings are made from the equivalent of medium grey cast iron, GG 25; although a pearlitic alloy, or ductile, spheroidal graphite, grade could be chosen, as discussed in [Chapter 2](#). Whilst this might appear an unlikely material choice, it has long been universally adopted where sliding metal seals are required, from small-bore air rifles to very

large diesel engines, either as plain rings or with special surface coatings to enhance wear in the most severe applications.

The magneto cover shown in [Photo 44](#) is a simple material selection exercise, the criterion being rapid casting into a die of a light material that has no strength requirement – it is merely a seal to keep the unit clean. This is a zinc (Zn) based die-casting alloy, needing only drilling of the fixing holes after casting. The carburettor ([Photo 45](#)), however, is far more complex in design as it has to incorporate fine internal passages into which removable brass jets are fitted, plus fuel regulation mechanisms and sliding throttle arrangement, whilst being compact, light, robust and rapidly produced from the similar Zn-base die-casting material, for its main body. In detail, there are a number of other features and materials, but as a proprietary item these are of less import here.



Photo 43 Piston with gudgeon pin and compression rings. Scoring shows prior minor seizure.



Photo 44 Zinc based alloy, fine die cast for non-structural magneto cover.



Photo 45 Carburettor assembly, fine die cast detail, containing complex internal shapes.

CASE STUDY 4: THE UBIQUITY OF MILD STEEL

These case studies are concluded with a couple of illustrations of the application of that most versatile of engineering materials, examined in more detail in [Chapter 2](#). One very important feature of mild steel is that, due to its widespread application over more than a hundred years, it is widely available from recycled products – large and small, industrial and domestic. This is complemented by the availability of new raw material, albeit at much greater cost than the former source.

As a brief introduction, a couple of diverse subjects are illustrated. The first is an example of non-functional artefacts that can be

produced purely for fun with mild steel in the form of 'horseshoe nails' (see [Appendix II: Resources](#)), welding rod wire (1.5mm dia.), recycled covered welding electrode cores (3mm, 4mm and 5mm dia.), and various other mild steel scrap from bolts and sheet metal. Occasionally, items such as small chain links or allen screws, though not strictly mild steel, can also be employed as imagination dictates.

These items have been constructed by the delicate application of oxyacetylene bronze welding from the small, shed-sized gas bottles (see [Appendix II: Resources](#)). A no.1 or no.2 gas nozzle with neutral flame, 'Easy-flo' brazing powder, and a no.1 'Sifbronze' rod provide the method shown in [Photos 47, 48 and 49](#); the result is shown in [Photos 50, 51 and 52](#).

The final example concerns the frames manufactured from scratch or heavily modified 'in the shed' for the increasingly popular 'pre-65' class of motorcycle trials bikes. This activity may be partly a nostalgic desire to escape from the modern variants, which are remarkably developed at the expense of taking manufacture beyond the shed, having attained a design convergence and sophistication familiar in all types of modern vehicles. Whilst highly impressive, this sophistication can inhibit true innovation as what we are led to believe is the 'ultimate product' is apparently reached.

[Photo 53](#) shows the JCS (Jim Caunt Special), shed built in 1964 as a highly innovative design compared to the conventional machine of its era; in fact it pre-dates much later designs that the corporate manufacturers eventually followed (although by that time the UK industry had vanished). The frame comprises a 4in/100mm dia. mild steel tube as the structural backbone and fuel tank from which the engine and suspension are attached, in a significant departure from the practice of the day. Fusion-welded using MMA (manual metal arc, otherwise known as 'stick welding'), it completed unscathed the punishing 800 road and moorland miles plus the long rocky sections of the international Scottish Six Days Trial of 1969, before being pressed back into active service since the 2010 season, in UK Peak District pre-65 trials.



Photo 46 Hot working steel, aka blacksmithing.



Photo 47 Oxyacetylene, versatile general heat source for hot working, welding and brazing/bronze welding.

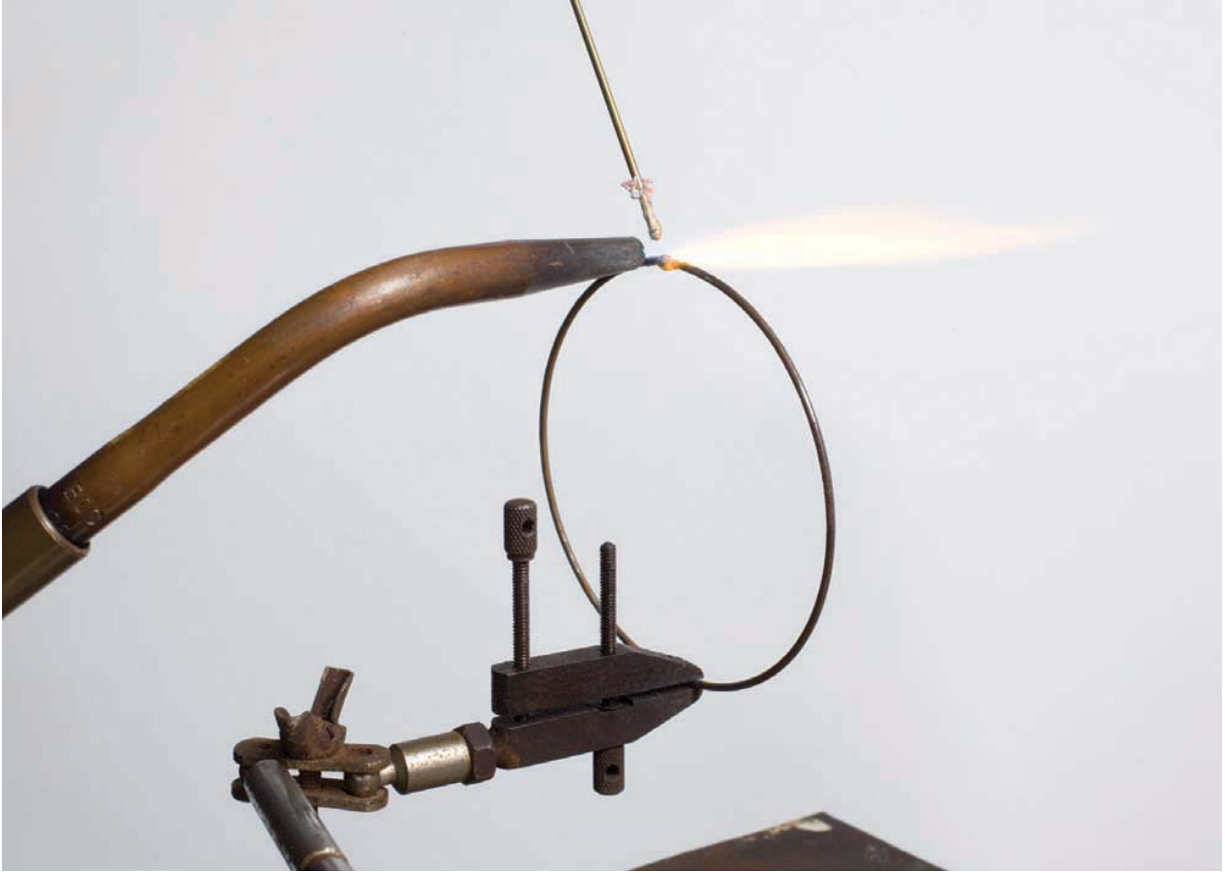


Photo 48 Bronze welding ring of $\frac{1}{16}$ " (1.6mm) mild steel welding wire as Photo 50.



Photo 49 Further bronze welding as Photo 50, small jigs and clamps required to hold components.

[Photo 54](#) completes this section with a BSA Bantam frame. The 1951 original was cut, heated and bent to the desired shape and new tubing bronze welded to cope with the demands of contemporary events. The forgiving nature of mild steel makes it a material well suited to such experimental work. A more sophisticated solution, possibly based on an experimental stress analysis as described in [Chapter 3](#), and use of the high-strength tubing from [Chapter 2](#), would be the logical next stage for optimizing strength and stiffness against weight.



Photo 50 Nailman model from mild steel nails, wire and sheet scrap.



Photo 51 Nailman Photore, nails, rod and mild steel sheet from scrap white goods.



Photo 52 Model, aka 1935 Brough Superior, from bolts, wire, chain link, reclaimed MMA electrodes and mild steel bar, from scrap stock.



Photo 53 JCS, pre-65 trials special, based on 4" (100mm) dia. mild steel tubing for frame and fuel tank.



Photo 54 1951 BSA Bantam based pre-65 trials, experimental project.

2 *Materials*

In our engineering projects, and with products in general, we need to view materials at a *macro* and *micro* level. In this chapter we consider a combination of the two (see also [Chapters 3](#) and [4](#)) so that we can explore materials that can be sourced for use in the home workshop.

The macro-scale is typically the larger picture, often considered the province of the mechanical engineer, where materials are employed for the elements of machines or structures, and their properties, strength and elasticity for example, are used to fulfil the design and satisfy intended use, as in ‘the specification’. On the other hand, the micro-scale is considered by the metallurgist or materials engineer, when the intricacies of the actual material within the machine or structure are examined. As is common in life, the finer details of materials within a particular application can be critical to the project, sometimes revealing weakness – such as fatigue or stress corrosion cracking – unseen from the macro-view.

A holistic approach is advocated, as success in making things on the scale envisaged here – we are not planning to build a nuclear submarine, hopefully – is best tackled with as wide an understanding as possible as we are operating primarily on the basis of our own resources or, at best, within a small group of enthusiasts. It soon becomes obvious why it takes such sophistication in organization and teamwork to tackle projects at the cutting edge of twenty-first-century engineering, which can require enormous resources in technical and scientific knowledge, as well as vast practical facilities. Such is obviously required for building a commercial jet engine; nevertheless it is equally reasonable for the home engineer to tackle

a working model of a jet engine, albeit a dramatically simplified project, for the ultimate consequences of failure are worlds apart.

Commonly available metals fall naturally into two groups: *ferrous* (containing iron) and *non-ferrous*.

FERROUS METALS

This group may itself be subdivided into steels and cast irons.

Steels

This section concerns: plain carbon (C); medium C; low alloy; high alloy (i.e. stainless); and special (e.g. nickel-base) steels – typically the most important category for the home engineer. Some guidance is required at this stage as there is surely ‘too much information’ available when it comes to the lists of these steels from the various standards and producers worldwide. There are many thousands of different alloys of steel listed, as large organizations, generally based in the major industrial countries, have a vested interest in adding complexity. Indeed, these organizations often seem as interested in evolving a practical and useful system of steel designations as the multinational car makers are in, not unreasonably, producing vehicles that can be rebuilt by the public.

It is probably fair to say that the USA is the leader in producing a coherent set of standards for steels (ASTM or AISI), whilst the UK is committed to European standards which, while perhaps admirable for particular industrial applications, are something of a dead loss for the home engineer. So I propose to revert, by several generations of standards, to the good old ‘EN’ standards (from ‘Emergency Number’ not ‘Euronorm’), formerly established in the 1955 series in the UK and still commonly used by practitioners. Subsequent standards have, perhaps, added precision and range, but at the expense of brevity (unlike the USA standards, which have retained their basic four-digit classification). The ancient EN designations suit our purposes, providing the worker with a fair chance of getting to

grips with the appropriate steel, in a language still understood by all interested parties, at least in the UK. By all means use equivalent, newer, even Euronorm, standards if appropriate but beware of the chasm between the range of steels listed and those available to the home worker. One is soon reduced from many thousands to the handful of steels reasonably available – those which form the basis of the following list.

Plain (C and CMn) and low-alloy steels cover the most basic ‘dead-mild’ steels (EN1A, EN3A), through medium carbon steels (EN8), to the direct-hardening or surface-treatable, low alloy steels (EN19, 24, 36, 40B). Note that the numbers don’t necessarily refer to ‘higher’ grade steels in a logical progression, but with a manageable list this should not be a problem.

The fundamentals of the most important steels are explored in [Chapter 4](#), but for the moment we consider just an overview. This provides a means of selection for particular applications, the range of which is surprisingly large, reflecting the extremes of properties available – from ‘dead-mild’ steel, close to the wrought iron as employed since medieval times by the village blacksmith, to the very high-strength low alloy steels as used in the most demanding designs for highly stressed transmission gears in high performance engines. Careful selection is important, as this wide range of strength is also associated with a varying workability, toughness and joinability. Increasing strength is associated with reduced ductility and variations in other properties, as illustrated. Joinability is often the keystone in being able to effect a practical solution to an engineering task in an efficient and simple manner, so the limitations when steels are joined by fusion (welding) and non-fusion (soldering, brazing, diffusion, bonding, etc.) processes are explored in some detail.

Table 6 Plain C and low alloy steels.

Grade	C	Mn	Si	Cr	Ni	Mo	other	Hv	UTS (N/ mm ²)	0.2YP (N/ mm ²)	elong. (%)	notes
EN 1A	0.07 – 0.15	0.8 – 1.2					0.2 – 0.3 S	103	360	215	22	free cutting, avoid welding
EN 3A	0.25	1						126 – 179	430	215	22	general purpose mild steel
EN 8	0.35 – 0.45	0.6 – 1.0						152 – 255	550 – 850	280 – 490	16 – 12	hardenable medium C steel
EN 16	0.3 – 0.4	1.3 – 1.8				0.2 – 0.35		201 – 352	700 – 1150	480 – 850	17 – 9	general purpose low alloy steel
EN 19	0.35 – 0.45	0.5 – 0.8		0.9 – 1.5		0.2 – 0.4		201 – 375	700 – 1225	480 – 940	15 – 8	high strength use
EN 24	0.35 – 0.45	0.4 – 0.7		0.9 – 1.4	1.3 – 1.8	0.2 – 0.35		248 – 444	850 – 1550	635 – 1235	13 – 3	higher alloy & strength
EN 40B	0.2 – 0.3	0.4 – 0.6		2.9 – 3.5	0.4	0.4 – 0.7		248 – 331	850 – 1075	635 – 740	13 – 9	for Nitriding
Ball bearing steel, EN 31A, AISI 52100	0.95 – 1.1	0.25	0.15 – 0.25	1.3 – 1.6				700 – 900	2300	2000		widely available, high accuracy, low cost
Silver steel	0.95 – 1.25	0.25 – 0.45	0.4	0.35 – 0.45				750 – 850				widely available, fine ground bar
Gauge plate (O1)	0.9	1.25	0.5	0.5			0.5 W	650 – 800				similar to Silver steel, in ground plate
Piano wire	0.55 – 1.1	0.25		1.3 – 1.6				700 – 900	2300	2000		widely available, very high strength, thin wire only, low cost

Plain (C and CMn) steels

For simplicity we can consider these as EN1A and EN3A, containing less than 0.25 per cent C and 0.7 per cent Mn. These steels are the most forgiving, able to accept cold working, with good machinability – a pleasure for ‘turning’ in the lathe, drilling or milling. EN1A is a free cutting steel, not advisable for welding.

In joining these steels by fusion welding, we are most likely concerned with: oxyacetylene (‘gas’); manual arc (‘stick’); metal inert gas (‘MIG’); or, for the aristocrat, tungsten inert gas (‘TIG’). These steels do not suffer from the heat affected zone (HAZ) problems of high local hardness that afflict the ‘higher’ grade steels, due to the lack of hardenability, which is one reason for their choice.

Surprisingly, even in our low C steels it is still possible to obtain some 'highly localized' significant increases in hardness, well beyond that which might be expected, as it is not normally possible to harden these steels by simple heat treatment with the usual rapid cooling, quenching, from $\sim 800^{\circ}\text{C}$. This is caused, self-evidently, by the far higher temperatures in welding than heat treatment, of $\sim 1500^{\circ}\text{C}$ at the 'fusion line' between weld and base metal. Such extremes provide some exceptionally high localized cooling rates and thus higher than expected spot hardness. As this effect is only produced over very small distances from the fusion line ($\sim 1\text{mm}$), and the values are still below those damaging to toughness and ductility, in all but the most sensitive industrial applications (such as in the petro-chemical industries handling 'sour gas') it can normally be neglected.

For the usual applications these steels can be happily beaten and welded without damage unless abused by excessive holding at high temperatures ($\sim 800^{\circ}\text{C}$) which can eventually lead to surface loss of carbon ('decarburization'), excessive surface oxidation, and grain growth. The lower-temperature processes such as brazing, bronze welding (simply a variation of brazing), and hard- and soft-soldering, are even more forgiving, but require good joint design to effect the required strength. Traditional cycle-lug designs are highly successful in employing soldering to make strong and light joints in frame tubes; but for simple 'butt' joints, fusion processes are required to achieve *full strength* connections (see [Table 23](#)).

It is thus apparent why 'mild steel' should be a favourite with the model engineer as it clearly lends itself to a wide range of mechanisms and structures, and is the material of choice for stands or jigs from angle, flat or tubular sections. It is cheap and readily available from scrap yards (with care in selection) or steel stockholders – at least the more enlightened who are prepared to deal with small order value, non-industrial, customers.

However, this panoply of admirable properties is limited where, for instance, greater strength is required. 'Rule-of-thumb' values are useful in determining if we have reached those limits. In Imperial

currency, take mild steel mechanical properties as 60 ksi for σ_u (UTS), 30 ksi for σ_y (0.2%PS), 30,000 ksi for E (Modulus of Elasticity), and 30 per cent for e (elongation). (For more details see [Chapter 3](#).) These properties are fine for simple structures like benches, frames and levers for 'classic' vehicles and many model applications. It is important, nevertheless, to recognize the limits where increased strength or fatigue resistance in shafts, as explored in [Chapter 5](#), is required for internal combustion engines or weight-sensitive structures like sports bicycle frames. As mild steel is not susceptible to heat treatment it must be avoided where highly stressed critical parts are encountered, such as a wheel spindle on a bicycle or motorcycle.

Whilst work-hardening increases the strength of mild steel, it also reduces ductility. Injudicious use of this can lead to local overloading and crack initiation, as will be seen when bending a heavy section of MS bar (given the strength, brute force, or power to do so!). High-strain forming should therefore be done at high temperature, $\sim 800^\circ\text{C}$, as 'cherry red'. This provides the means of forming mild steel into some wonderfully complex and delicate shapes, as practised by the village blacksmiths since ancient times.

Where only surface hardness is required, as is often the case in rotating joints like the crankpin of a model steam engine, mild steel is often used with 'case hardening' (see [Chapter 5](#)). Additional carbon is diffused into the surface to form a thin skin (typically 1 to 40 thou, 0.025 to 1.0 mm), which can be heat treated to a hard, wear-resistant surface that surrounds the relatively soft and tough mild steel core. A process that can be successfully tackled without industrial resources, it is an invaluable addition to our armoury of practical treatments.

Medium C steels

The natural progression from the mild steel (EN1A or EN3A) is to medium C steels containing 0.4 to 0.6 per cent C, with Mn 0.5 to 1.2 per cent, as typified by the well known EN8 or '40-carbon' steels. A

versatile alloy, when in softened condition, slow cooled from $\sim 850^{\circ}\text{C}$, it is comparable to mild steel. With judicious heat treatment, of which more in [Chapter 5](#), it is possible to achieve relatively high strengths of 100–120 ksi (700–850 N/mm) as ‘condition R’, but at the cost of ductility reduction to 18 per cent.

This sounds like the ideal partner to mild steel, but one particular snag is that these properties can only be achieved in relatively small sections due to the need for high cooling rates to achieve the required transformation to martensite, the hard phase responsible for the increase in strength.

Nevertheless, these medium carbon steels are a relatively simple alloy, in composition, if not detailed metallurgical characteristics like phase transformations, as explored in [Chapter 5](#). Particularly important is their availability in relatively small and varied sections, being a standard for ‘key steels’, the rectangular-section locating keys, for fixing pulleys to drive shafts. These are often classified as ‘C45’ or AISI 1040 grade, close to EN8.

Where fracture resistance is paramount, such as in the moderately stressed wheel spindle, EN8 is probably not the ideal choice and we may turn to the intrinsically tougher, lower carbon type, exemplified by EN16. This sacrifices some strength, due to its low alloying content, for ductility and toughness, whilst still possessing improved ‘through-thickness’ heat treatment capability. It provides a versatile general-purpose steel for a variety of uses where stresses are not too high, and conditions not too arduous. Limitations arise where strength becomes more critical, which also applies to fatigue-sensitive parts, like highly loaded shafts. For these, we need the more sophisticated low alloy steels. Fatigue limits are closely related to yield strength, when considering smoothly machined shafts, common in vehicle motors and transmissions. This is discussed further in [Chapter 4](#).

Low-alloy steels

We are now in serious material country, these alloys being the workhorses of modern machinery and vehicles, used where parts are really working hard, albeit normally unseen, as the remarkably reliable materials forming the gears and shafts transmitting the power of road vehicles and model internal combustion engines.

For our purposes, we do not need to be over-concerned with the intricacies of alloy selection that would be involved for use in several million passenger cars of a single component design, where a minute miscalculation could spell commercial disaster. We are principally concerned with the ('metallurgist's favourites') EN19 and EN24 types, with ~0.4 per cent C, 1 per cent Cr and varying Ni and Mo. These are highly versatile alloys with a wide range of strengths and ductility, by the well-established heat treatments of quenching and tempering. These alloys can achieve great strength without too severe a loss of ductility and toughness.

As applications become more critical it is important to consider 'toughness', the resistance to fracture when provoked by stress-complexity, surface imperfections, low temperatures, or combinations thereof. The optimum performance of these types of alloys may also be achieved through more sophisticated heat treatment regimes – bearing in mind that simple 'quench and temper' has been practised since the days of the Roman Empire. During the twentieth century the complexities of apparently simple steel alloy systems have been unravelled as investigations have progressed by direct observation down to the atomic level, revealing the way that the various phases, like martensite and the bainites, control the characteristic of fracture toughness.

We can now select methods of heat treatment within the grasp of the enthusiast, even austempering of low alloy and spring steels which, combined with some simple transformation diagrams (see 'TTT diagrams' in [Chapter 4](#)), can replicate industrial-style treatments. This mimics the methods used in vehicle component manufacture which produce such extraordinary reliability of hard-worked parts, like internal combustion engine valve springs.

Beyond EN19 and 24, the short-list includes those with greater Cr and/or Ni, such as 40B, and the ball-bearing steel AISI 52100

(similar to EN39). These alloys all have particular virtues, such as an ability to achieve specific mechanical properties in larger sections, but selection is to a degree a matter of personal preference. There is a good case to be made for extensive experience with a particular alloy, as the more use and knowledge gained the better. Some simple testing, starting with hardness, can be tackled by the home worker, to build up a store of knowledge and confidence for important applications that need to be made safe from failure.

Silver steel and gauge plate

These are important because of their ready availability. Supplied as precision-ground bar, silver steel makes a practical choice for the model engineer who requires a material superior to mild steel, but lacks the facilities or inclination to engage in heat treatment.

Silver steel is, of course, a misnomer based on its characteristic appearance as a bright, finely-ground bar. The high C, and minimal alloy content, provides potential for considerable hardenability, but it is normally supplied in the annealed (softened) condition. Whilst the high carbon content, ~1.2 per cent C, means it is more difficult to cut by hand and less amenable to a good machining finish than mild steel, it can still be worked satisfactorily in the as-received condition. The composition, with its potential for hardenability, is not always fully utilized, apart from when making tools and cutting bits, which is a useful feature indeed. Such uses can be obtained by simple methods of heat treatment, without recourse to elaborate heating and tempering facilities, when variations in hardness and loss of ductility present no practical problems.

Gauge plate is in a similar class to silver steel, but normally supplied as precision-ground flat plate or a rectangular bar instead of round sections. It can often be used in place of silver steel, as an alternative to mild steel. Its capability for hardening is valuable for making jigs and guides, where hardness is paramount, after cutting and machining to the finished size, and where ductility loss is acceptable.

Spring steels and piano wire

These are specialized materials, the principal requirement for spring steel being its ability to absorb elastic deformation without losing shape, beyond the elastic limit and into the 'plastic', non-recoverable, region. Therefore, a high yield strength (σ_y), and high 'modulus of elasticity' (E), is required. This is achieved particularly with the higher-strength steels like piano wire. In contrast, titanium or aluminium alloys are not preferred for springs, with their unfavourable E value, 50 per cent lower than steel.

The difficulty for the home worker is obtaining material in the preferred size, as round wire or thin strip. For springs comparable with those produced commercially, a high-strength alloy is required. This calls for annealing before forming to the required shape, then heat treatment, preferably by austempering or quench and temper to obtain full strength and service capability.

Piano wire is the strongest steel commonly available, albeit only as thin wire. It begins with a composition susceptible to heat treatment, of ~1 per cent C. When drawn through dies it is work-hardened, deforming the heat-treated microstructure, the process being repeated through several stages. This produces thin wires of exceptionally high strength, ~250 ksi/1750 N/mm². However, there are clearly limitations for the use of these wires – apart from springs, they can be found in multi-core tension-elements, as supporting cables for heavily loaded structures, for both full-size structures and models.

Stainless steel

This covers a wide range of high-alloy steels, with greater than ~10 per cent Cr conferring corrosion resistance. This is not, unfortunately, the case for every possible environment or application, so careful selection is required for critical applications. The main categories are outlined below, with a more detailed

understanding following from the basics of ‘equilibrium diagrams’ and microstructures described in [Chapter 4](#).

Austenitic stainless steels

The most familiar of the stainless family, and for our purposes, the most important category, typified by grade 304 (‘18/8’) and grade 316.

These alloys have a single-phase (‘austenite’) microstructure, providing relatively low yield strength and high ductility with a high propensity for work-hardening. They are very effective in resisting many common corrosion environments such as for household cutlery and domestic uses, and in the inclement conditions all too familiar in the UK! Particular weaknesses include applications in certain marine environments, when seaborne spray enters equipment such as highly-stressed boat fittings, for example – which came as something of an unexpected challenge when in August 1966 Francis Chichester (later knighted by the Queen with the same sword Elizabeth I used to enoble Sir Francis Drake after his first crewed circumnavigation in 1581) set sail from Plymouth. He soon found his brand new Gr. 316 self-steering gear rendered useless by stress corrosion cracking (SCC), a few days into the first solo circumnavigation of the world – a tough lesson in material selection! Nevertheless, this family of alloys, all variants on the same basic austenitic theme, are extremely useful not only for their normal corrosion resistance. Grade 316 is still widely used for human body implants, especially for pins and wires. Another important feature is their ability to perform well at the elevated temperatures that would immediately lead to severe scaling and oxidation in the plain or low-alloy steels.

Whilst there are intermediate temperatures where they should not be held for extended times, such as encountered in 475°C embrittlement, particular alloys such as Gr.321, ‘stabilized’ with the small addition of niobium (Nb) or titanium (Ti), can be safely used at surprisingly high temperatures, ~800°C, without significant degradation. This provides a very useful alternative to the expensive, and hard to obtain for the home engineer, Ni-base alloys

such as 'Inconel 625', when making high temperature parts for model jet engines.

Table 7 Austenitic stainless steels.

Grade	C	Cr	Ni	Mo	Si	Mn	Ti	Nb	Other
301	0.15	16 – 18	6 – 8		1	2			
302	0.15	17 – 19	8 – 10		1	2			
303*	0.10	17 – 19	8 – 10		1	2			0.15 S 0.45 P
304	0.08	17 – 20	8 – 10		1	2			
310	0.025	24 – 26	19 – 22		1.5	2			
316	0.08	16 – 18	10 – 14	2 – 3	1	2			
316L	0.03	16 – 18	10 – 14	2 – 3	1	2			
321	0.08	17 – 19	9 – 12		1	2	Min 5 × C		
347	0.08	17 – 19	9 – 13		1	2		Min 10 × C	
254 SMO	0.02	18 – 20	17 – 19	6 – 7	0.8	2	1.9	0.2	0.3 Al

*303 is used for its free-machining characteristics useful for making nuts and bolts; but not recommended for welding. Corrosion resistance still OK in mild environments. Good availability.

There was a time, long ago but still erroneously regarded as a problem by some, when loss of corrosion resistance as 'weld decay' was experienced when welding these austenitic stainless steels. With the widespread use of ultra-low carbon alloys (designated, for instance, 304L or 316L), coupled with welding electrodes stabilized with Nb or Ti, the phenomenon of weld decay should not be encountered. It was caused by the reaction of the parent metal close to the heat affected zone (HAZ) being subjected to a critical range of temperature that enabled C precipitation to form ' $C_{r23}C_6$ carbides', so depleting this zone of Cr and thus reducing corrosion resistance adjacent to the weld. When welding these alloys it is, therefore, important to ensure that low carbon (<0.04 per cent C) alloy is used, or exercise alternative strategies such as special electrodes or avoidance of particular applications.

Another feature is that these alloys cannot be hardened by heat treatment, but possess high rates of work-hardening, the only means of producing through-thickness strengthening. Conversely, to

remove the effects of work hardening, heat treatment at $\sim 1100^{\circ}\text{C}$ is required, followed by a rapid cooling, preferably water quenching to arrest the high-temperature austenitic phase. This high temperature heat treatment, annealing, *not* hardening, normally makes it impractical to stress-relieve welded structures, as is often done for the CMn steels which require only 600°C , held for an hour-per-inch of thickness.

Martensitic stainless steels

These alloys contain typically 12 per cent Cr and higher C than the austenitics, but negligible Ni, and form a martensitic structure at RT (room temperature) without recourse to special heat treatment. Popular alloys are Gr. 410 and 420 providing higher strength than the austenitics, but requiring a tempering treatment for reasonable ductility. Corrosion performance is inferior in many common environments but improved in others. They are therefore useful for more highly stressed parts like gears and fasteners or for high temperature use as turbine blades. The higher Cr and C versions (e.g. 431 and the 440s) are sufficiently strong and hard for applications in ball bearings and moulds, where the need for corrosion prevention precludes the low-alloy steel alternatives.

This hard martensitic structure means that they present more problems than the austenitics for welding and special precautions (such as high pre-heat, to reduce cooling rate) are needed to avoid an unacceptably hard and low toughness HAZ. Heat treatment after welding is not normally practical, requiring over 1000°C holding temperatures, prior to quenching and tempering.

Table 8 Martensitic stainless steels.

Grade	C	Cr	Ni	Mo	Si	Mn	P	S
410	0.15	11.5 – 13			0.5	1	0.04	0.03
416	0.15	12 – 14		0.6	1	1.25	0.04	0.15
420	0.15 – 0.4	12 – 14		1	1		0.04	0.03
431	0.2	15 – 17		1.3 – 2.0	1	1	0.04	0.03
440A	0.06 – 0.08	16 – 18		0.8	1	1	0.04	0.03

Ferritic stainless steels

A relatively small compositional variation with higher Cr and lower C stabilizes the ferrite at all temperatures. This renders these alloys non-heat treatable, with lower strength but higher ductility than the martensitic types. The lean composition makes for a lower cost alloy, which finds a variety of applications, including decorative trimmings.

Table 9 Ferritic stainless steels.

Grade	C	Cr	Ni	Mo	Si	Mn	S	P	other
405	0.08	11.5 – 14.5			1	1	0.04	0.03	0.1 – 0.3 Al
409	0.08	10.5 – 11.75			1	1	0.05	0.05	Ti = 6 × C
429	0.12	14 – 16			1	1	0.04	0.03	
430	0.12	16 – 18			1	1	0.04	0.03	

Duplex stainless steels

These are a relatively recent development, with a microstructure of around 50/50 austenite and ferrite, with 25 per cent Cr-5 per cent Ni as the compositional basis. Developed initially for high integrity equipment in offshore oilfield exploration, they provide high strength and toughness, with a corrosion resistance in seawater that has seen their widespread use as an alternative to the expensive high Ni alloys.

A price is paid when welding, however, as the phase balance is easily disturbed and with it the loss of many of the outstanding properties available from the fully heat treated material, water quenched from $\sim 1050^{\circ}\text{C}$. It is possible to successfully weld duplex, but only with appropriate skill and knowledge. However, it is unlikely that use of this material will be essential for the home engineer in the near future.

Table 10 Duplex stainless steels.

Grade	C	Cr	Ni	Mo	Cu	N	W	PRE = C + 3.3Mo + 16N (pitting resistance no.)
UNS S32900; 329	0.08	23 – 28	2.5 – 5.0	1 – 2				26
UNS S31800; 2205	0.03	22 – 23	4.5 – 6.5	3.0 – 3.5		0.14 – 0.2		34
UNS S32760; Zeron 100	0.03	24 – 26	6 – 8	3 – 4	0.5 – 1	0.2 – 0.3	0.5 – 1	40

Other stainless steels

Beyond the stainless steels noted above there are several others, essentially variants, such as the ‘super-austenitics’ like ‘SMO 254’, which are designed to overcome particular weaknesses such as in marine applications. There are also special high-strength alloys like the ‘17.4 PH’ (‘precipitation hardened’) which may be worthy of consideration for special applications.

Cast iron (CI)

Whilst there is a great variety of cast irons (grey, white, nodular, ductile, malleable, whiteheart, blackheart, pearlitic, malleable, etc.), the ones of most relevance are grey, white and nodular – a superficial knowledge of the remainder of the cast iron family is sufficient for our purposes. An understanding of this commonly available but metallurgically complex material demands reference to equilibrium diagrams and microstructures as described in [Chapter 5](#), but here we are simply concerned with an overview and practical applications.

The useful properties that make CI such an important material include:

- ◆ Ease of casting (fluidity) and ability to provide good pattern replication.
- ◆ Relatively low melting temperature ($\sim 1200^{\circ}\text{C}$), significantly lower than steels.
- ◆ Ease of machining (for particular grades).
- ◆ Range of properties from low- to high-hardness; good compressive strength; ductile grades; and range of corrosion resistance; over wide range of costs, from very low upwards.
- ◆ Good availability, for certain types.

The principal feature of cast iron is its higher C content, usually >2 per cent C, an order of magnitude greater than for mild steel. The C exists in cast iron either as free-graphite (normally as 'flakes' unless modified) or combined with ferrite to form 'cementite', a white, hard and brittle compound. Flake graphite is found in the grey irons and cementite in white cast irons, relating to the characteristic appearance of their respective fracture faces.

Grey CI

A critical aspect for producing simple grey cast iron is the cooling rate. A low rate favours grey cast iron, where the excess C is ejected from the melt as free-graphite in the form of long flakes. This produces a matrix of relatively soft ferrite and a material of low hardness (150 H_V) and strength approaching mild steel. Although satisfactory under loading in compression, as normal in machine-bed or support blocks, the microstructure containing graphite flakes is extremely damaging in tensile loading. It provides virtually no ductility (i.e. extension beyond yield prior to failure). This contrasts dramatically with the ductility of mild steel. The lack of ductility caused the Victorians much grief when applied to vessels or structures subject to tensile loading, sometimes unexpectedly, and

contributed to many disasters. Where used judiciously, however, as in the world's first all-metal bridge at Ironbridge, high above the River Severn (Shropshire, UK), which is still standing, it is a testament to the capability of an early cast-iron structure where critical tensile loading is avoided.

White CI

This is produced where composition and cooling rate combine to produce graphite in the form of cementite, a notoriously hard and brittle phase. It finds many uses where such properties are beneficial, as in hand files and cutting tools of appropriate design. Indeed, such is the importance of cooling rate in castings that, as a function of thickness, it can often vary in a single casting and produce a range of microstructures from grey to white.

Clearly, it is important when machining cast iron to understand something of the expected microstructure. A fully hard white cast iron is not amenable to conventional machining and grinding is necessary.

Silicon is a vital element in cast-iron manufacture, de-stabilizing the cementite and promoting a grey cast-iron structure. Sulphur (S), manganese (Mn) and phosphorus (P) are all heavily involved in these cast irons. Sulphur opposes the effect of silicon, inhibiting graphitization and tending to produce brittleness, when forming iron sulphide, (Fe-S).

Manganese can form Mn-S, a sulphide insoluble in Fe. It therefore promotes graphitization but has a stabilizing effect on carbides and generally acts to increase cast-iron hardness and promote grain refinement, thus favouring an increase in strength.

Phosphorus is present as the phosphide Fe_3P , and the eutectic formed melts at 950°C , providing excellent fluidity for high-phosphorus cast irons. Whilst very useful for fluidity in thin sections, it also confers high hardness and lack of ductility, leading to particularly brittle cast iron, if not carefully controlled.

Nodular (ductile) CI

The graphite flakes responsible for the lack of ductility can, almost magically, be transformed by a process of 'spherodization', turning the linear flakes into compact balls and thus obviating their chronic stress-raising effect. This is achieved by a well-understood process usually involving a timely inoculation of, typically, cerium (Ce) or magnesium (Mg) just prior to casting. The effect is dramatic – and easily seen in the microstructure – changing the cast iron from a material with virtually no ductility in tension, towards that comparable to a respectable steel of similar strength. Precautions are required, such as providing >2.5 per cent Si, to prevent white iron formation in thin sections, where Mg is used as an inoculant. But there are other variations possible, and more recent developments with heat treatments like 'austempering' have been used extensively to provide ductile castings for the most rigorous components. Austempered ductile iron suspension links even compete with welded steel structures, for use in volume-production cars (see [Chapter 5](#)).

Alloy CIs

The range of specialist cast irons developed for particular applications is extensive, with additions such as Ni, Cr, Mo, V and Cu, to enhance corrosion resistance, reduce section sensitivity, or enhance particular mechanical properties. Martensitic cast iron can be designed to provide appropriate structures by varying heat treatments, and austenitic cast iron, with 10–20 per cent Ni, and 5 per cent Cr – 'Ni-resist' – provides corrosion resistance and non-magnetic castings.

Malleable CIs

Processes such as 'Blackheart' and 'Whiteheart', for malleable CI, were once widespread and appear in the literature but have largely

been superseded by the ductile CI family of CIs and the alloy CIs described above. As with the specialist developments in stainless steels, a profusion of alloys has been developed in response to commercial opportunities which may, or may not, be useful for our purposes.

NON-FERROUS METALS

Copper and copper-base alloys

Next in importance and usefulness to the ferrous alloys, these provide advantages in corrosion resistance and electrical conductivity which the ferrous cannot match. A wide range of alloys has been developed from very early times, for copper ore was the first metal found by man in pre-historic times, where the low melting point for pure Cu, at 1083°C, and nearer to 800°C for the bronzes, has enabled their casting ever since the Bronze Age.

Copper

Copper (Cu) is unusual as it is one of the few elements that are used in engineering in their pure, or almost pure, state. A low strength makes it suitable only where stresses can be limited far below those taken for granted in the ferrous alloys. Remarkable for its electrical and thermal conductivity, combined with corrosion resistance, it is the undisputed material of choice for the boilers of model steam engines and locomotives, as seen in [Case Study 1 \(Chapter 1\)](#). This is despite its relatively high cost, an example where the properties of the material outweigh such considerations; a lower-cost alternative, like mild steel, would severely compromise design.

The excellent electrical conductivity makes Cu the most common material for electrical work, from small- to large-scale wiring and components, accounting for >50 per cent of Cu production. ETP (electrolytic tough pitch) Cu is not preferred for welding and contains

more oxygen than OFHC (oxygen-free high conductivity) Cu, the latter having superior electrical conductivity. Low strengths can be increased by cold-working, at the cost of a slight loss in conductivity. The significant loss of strength above $\sim 200^{\circ}\text{C}$ can be compensated for by additions of silver (Ag), permitting use up to 340°C for applications such as turbo-generators and higher-temperature electrical motors.



Photo 55 Model locomotive boiler, firebox, copper fabrication.



Photo 56 View from boiler cylinder of Photo 55.

Above 340°C, ETP-Cu can suffer from hydrogen embrittlement when used in a reducing atmosphere, so oxygen-free or phosphor-deoxidized Cu is required. The former is noted for its ductility, and finds applications in severe deep-drawing and cold-working operations.

Items requiring greater strength, approaching that of mild steel (as for seam-welding rolls), require heat-treated Cu with Cr additions but

at the cost of ~20 per cent reduction in conductivity. These alloys are also commonly used in high-volume production of vehicle radiators and industrial heat exchangers.

Cu is also familiar in domestic water systems, making use of its thermal conductivity and corrosion resistance, with self-evident success, providing a useful source of material for our purposes.

Copper alloys

The major categories are the brasses (Cu-Zn), bronzes (Cu-Sn), and Cu-Ni alloys. These have undergone substantial development over the ages, providing materials suitable for casting, and wrought forms of significant interest to the home engineer.

Brasses

Whilst these alloys are essentially a mixture of Cu and Zn, they commonly also contain small proportions of Sn, Pb or Al. In order to understand the different characteristics that occur as a consequence of varying amounts of Zn we need to skip forward to the equilibrium diagrams described in [Chapter 5](#). This provides an illustration of the different phases, with the ductile 'α-phase', an 'fcc' (face-centred cubic) atomic structure, occurring in the range from 0 to 37 per cent Zn. These 'alpha brasses' behave as would be expected, responding readily to cold-work, enabling substantial work-hardening before the onset of failure.

As Zn is increased from 37 per cent to 55 per cent, then 'β phase', or combinations of α and β phase, result. As a 'bcc' (body-centred cubic) structure, β phase has different characteristics from alpha brass and is known, unsurprisingly, as 'beta brass' or α/β brass. These possess higher strengths with lower ductility than the brasses, and are susceptible to heat treatment. Working is preferably done at high temperature, where the composition is fully β phase. Rapid cooling from the β phase region can produce a distorted structure, known as β', a martensitic phase, although not the hard and brittle transformation product characteristic of steel

martensite. This phase can also, as seen in [Figure 62](#), be produced at slow cooling rates for certain compositions.

Further increase in Zn, above 55 per cent, produces γ phase, notably brittle and thus well avoided. Compared to Cu, brasses possess higher strength and reduced cost, combined with good machinability, general resistance to atmospheric corrosion, and good thermal and electrical conductivity. As seen above, their strength increases up to 40 per cent Zn, the familiar '60/40 brass', although with some loss in electrical and thermal conductivity. The brasses provide a good compromise between strength and ductility, and are chosen where the material is to be cold worked. However, the high toughness of these brasses is not ideal for machinability and small additions of Pb (<4 per cent) are used. These alloys are noted for their aesthetically attractive colouring, yellow to red, depending on their composition.

Table 11 Brass.

Type	Cu	Zn	Pb	UTS (N/mm ²)	Elongation (%)	Typical applications
Gilding Metal	95	5		250	43	coins, medals, base for gold- plating
90/10 Bronze	90	10		270	44	marine hardware, jewellery
Red Brass	85	15		285	45	heat exchangers, plumbing
Cartridge Brass	70	30		335	60	springs, rivets, radiator cores
Yellow Brass	65	35		350	60	plumbing accessories
Muntz Metal	60	40		375	45	hot forgings, tubing
Leaded Bronze	89	9.25	1.75	255	45	screws, builders materials

Corrosion resistance: brasses

Whilst the brasses generally possess good resistance to atmospheric corrosion in domestic, industrial, and marine environments, there are circumstances where they suffer significant damage, commonly due to corrosion mechanisms such as 'season cracking' and 'de-zincification'.

Season cracking is readily observed, resulting from a susceptibility to intergranular or stress corrosion cracking in certain environments. The solution is to relieve the internal residual stresses induced by cold working during manufacture. Annealing by heating at $\sim 280^{\circ}\text{C}$ can achieve sufficient stress-relief and should be utilized where season cracking could be experienced.

De-zincification can afflict some α - β brasses in sea water, or fresh water in high O_2 or CO_2 concentrations. This can produce a pitting-corrosion attack, with penetration of porous Cu surface layers, leading to leakage through these layers. The effect can be

minimized by addition of small amounts of tin (Sn) and antimony (Sb).

Bronzes

Tin is the principal element added to Cu to produce bronze. In a manner similar to that used to describe the brasses, we refer to the equilibrium diagram of Cu-Sn, a 'peritectic' like [Figure 61](#) with α , β and γ phases to ~30 per cent Sn. These tin-bronzes are complemented by the aluminium bronzes, silicon bronzes, and beryllium bronzes, all being of commercial importance ([Table 12](#)).

Table 12 Bronze.

Type	Cu	Sn	P	Other	UTS (N/mm ²)	Elongation (%)	Typical applications
95-5 Phosphor Bronze	94.8	5	0.2		325	64	bolts, electrical contacts
90-10 Phosphor Bronze	89.8	10	0.2		450	68	gears, bushes, marine hardware
Gun Metal (sand cast)	88	10		2 Zn	275	30	gears, bearings
Aluminium Bronze (heat treated)	86			10.5 Al, 3.5 Fe	685	12	higher strength parts, couplings
Beryllium Bronze (heat treated)	98			1.7 Be, 0.3 Co	1175	5	ultra high strength parts, springs

Tin bronzes

The tin bronzes will form a combination of phases unless rapidly cooled from the single phase region, such as alpha (α) phase at <15 per cent Sn. Whilst strength increases with Sn addition, higher amounts of Sn can produce a hard Cu_3Sn compound, which reduces ductility. As tin bronzes contain up to 0.3 per cent P, acting as a deoxidizer (removing O_2) and improving the fluidity of castings, these are also known as 'phosphor bronze'. This P addition also increases strength, with the formation of the compound Cu_3P .

Aluminium (Al) bronzes

The addition of Al is shown in the equilibrium diagram ([Figure 64](#)) but this is only a general guide as small additions of Fe, Ni, Mn and Sn will modify this diagram considerably. The important range is from 4

to 11 per cent Al, providing alloys that find application in critical products where high strength, approaching that of mild steel, is combined with excellent corrosion resistance in seawater. These alloys are widely used for ship propellers, valves and many marine components.



Photo 57 Early vehicle bulb horn, c. 1920, for vintage motorcycle.



Photo 58 Detail of Photo 57, showing characteristic brass season cracking.

These are not particularly easy materials to cast but, because of their commercial importance, methods for satisfactory production have been developed. The pioneering work of Henri Durville (1818–81) in France was followed by that of Charles Meigh (1892–1968) in England, who perfected the Durville process for casting. Meigh Castings Ltd have continued the development of industrial and marine products to the present day, and a wealth of knowledge is available for critical applications.

Beryllium (Be) bronzes, whilst of higher strength with up to 2.5 per cent Be, are more specialized and expensive, and unlikely to be of particular importance to the small-scale operator.

The use of bronze includes such diverse products as springs, bearings, marine fittings, and castings for bells and statues. The aluminium bronzes, in particular, provide an attractive golden appearance, resistant to atmospheric degradation. Church bells obtain their beautiful tone at a Cu-20 per cent Sn composition, whilst

bronze statues often add Pb to around Cu-15 per cent Sn for a characteristic grey-black patina.

Copper-nickel alloys

These are notable for their particularly high corrosion resistance and cost, finding industrial application in severe environments. Cu with up to 30 per cent Ni forms a single phase structure, not susceptible to heat treatment, but with high ductility, making it capable of work-hardening. These alloys are commonly known as ‘cupronickels’.

Alternative alloys based on Cu-Ag are used for fine-scale equipment, including optical and camera parts. Finally, ‘Constantan’ is a widely used alloy with particular electrical characteristics employed in temperature-measuring equipment ([Table 13](#)).

Table 13 Copper–nickel.

Type	Cu	Ni	Zn	Mn	UTS (N/mm ²)	Elongation (%)	Typical applications
Cupro-Nickel	70	30	0.2		345	50	surgical instruments, marine, heat exchangers
Nickel-Silver	65	12	23		390	43	instruments, optical
Constantan I	58.5	40		1.5			temperature measurement, thermocouples

The light metals

These, third in our most important category of metals, include aluminium and its alloys, magnesium, and titanium. Whilst delivering great value from their low density, some also have valuable corrosion resistant, electrical and thermal-conductivity properties. The important aluminium castings are usually based on Al-12 per cent Si alloys, as below.

The relative densities are: Al 2.70 g/cm³, Mg 1.74 g/cm³, Ti 4.54 g/cm³ – compared to typical values for steel of 7.9 g/cm³, and 8.7 g/cm³ for Ni. Thus, as a rule of thumb, mild steel is three times heavier than Al, 4.5 times heavier than Mg, and 1.75 times that of Ti. This potential advantage is only realized if yield strength (σ_y) and stiffness (E) are comparable – and thereby hangs the tale, as related next.

Aluminium (Al)

Pure aluminium as reduced from its ore, bauxite, is an abundant element and despite the cost of processing, largely due to the energy required for refining, is relatively cheap, plentiful and available to the home engineer in a variety of forms – from domestic cooking foil to large castings. It is endowed with excellent corrosion resistance due to its thin but tenacious passive oxide film, which is readily enhanced by ‘anodizing’; its electrical conductivity, 60 per cent of the conductivity of Cu; an excellent workability; and machinability. It is, nevertheless, restricted in engineering applications due to its poor strength and hardness.

Where this strength deficiency is not important, as in lightly-loaded parts, it provides a delightful material with which to work, its ductility permitting severe forming and easy metal removal in the small lathe. Its non-toxic, non-magnetic and non-sparking characteristics can also provide solutions to particular problems.

However, as will be examined further in [Chapter 5](#), the importance of Al lies in its facility to overcome the strength deficit through one or more of the following: work-hardening (a two-fold increase in strength); adding alloy elements (e.g. Cu, Mg, Si – providing a four-fold increase); heat-treating (alloys containing Cu, Mg and Zn); and dispersion hardening (with hard, insoluble, aluminium-oxide (Al_2O_3) particles). Further possibilities for industrial use, if not for the home worker, include aluminium used as a matrix into which strong fibres are incorporated, for ultra-high strength composite materials, albeit at very high cost.

Aluminium alloys

Aluminium castings

With a melting point between 660°C and 577°C, at the eutectic point for the Al-Si alloys, these alloys present practical possibilities for casting in the home workshop, with optimum fluidity around 11 to 12 per cent Si, as per the popular alloy LM6 (seen in [Chapter 1](#), Case

Studies 2 and 3). The equilibrium diagram ([Figure 63](#)) shows the region of interest, from 1 per cent to 12 per cent, but when cooling in a sand mould such alloys produce an acicular microstructure, for example, Al-12 per cent Si, resulting in poor ductility. However, in commercial foundries it is possible to regain ductility in a manner somewhat analogous to inoculation in grey cast irons. A small amount (~0.01 per cent) of sodium can be injected into the melt prior to pouring, effecting a 'modification' that transforms the microstructure and properties.

Modified Al-Si alloys, as in the 12 per cent Si casting example, can achieve improvements in tensile strength and elongation, from ~7.5 ksi (120 N/mm²) to 30 ksi (200 N/mm²); and from 5 per cent to 15 per cent, respectively. It also makes possible an increase in Si, from the eutectic value of 11 per cent to 14 per cent, and at an even lower temperature. An increased cooling rate can also refine the unmodified Al-Si structure, but this is often difficult to effect in practice. Provided that the improved structure is achieved, the basic Al-Si, often with small additions of Mg, Zn and Cu, provides an excellent casting alloy, which benefits from intrinsic high fluidity, and enables reproduction of intricate shapes, impractical to produce by machining. Typical examples include model, and full size, cylinder heads and barrels with extensive finning, familiar in air-cooled internal combustion engines. The high thermal conductivity and light weight makes this a highly popular material choice, especially practical with its low casting temperature.

The alloys accounting for the majority of castings in the UK are based on the LM2, LM4, LM6 and LM25 grades. The binary Al-Si alloys are *not* susceptible to heat treatment for increasing strength – but those with significant alloying elements *are* heat treatable by the precipitation hardening process, quenching from high temperature then 'ageing' at the appropriate temperature, usually between room temperature and 200°C.

Wrought alloys

Whilst the benefits of the *cast* alloys are considerable, nevertheless where optimum mechanical properties are required, such as aircraft structures or critical parts of vehicles, we turn to the *wrought* alloys. The desired shapes must be produced by machining from solid, or joining by mechanical means, adhesives, or fusion welding (for the weldable Al alloys). The wrought alloys lend themselves to easy machining and shaping, within the limits of their ductility.

Classification is most sensibly approached by the international and BS standards four-digit system: 1000 (Al); 2000 (Al-Cu); 3000 (Al-Mg); 4000 (Al-Si); 5000 (Al-Mg); 6000 (Al-Mg-Si) and 7000 (Al-Zn) series. However, it is also common for manufacturers' and local standards to be used, without reference to the fundamentals which provide a more rational means of wrought alloy selection. The home worker rarely has the luxury of selecting an optimum grade, so usually has to take what is available and then assess its suitability for the project, rather than the ideal of deciding on a specification and ordering it up! Aluminium alloys seem to be particularly afflicted by this market confusion for materials available to the home engineer. Amusingly, even some very experienced workers will refer to these aluminium alloys simply as 'alloy' material – stemming from a confusion between 'ally' (implying pure Al), and 'alloy' (the eponymous alloy) – bearing in mind that almost all practical metals are, of course, alloys.

Often, the excellent properties of the wrought Al alloys need to be balanced with a keen knowledge of joining limitations. It is necessary to examine the grades discussed above, unless we are unconcerned about mechanical properties. In [Chapter 5](#) the mechanisms of strengthening by the long-established precipitation-hardening ('ageing') process are outlined, but it is perfectly satisfactory to employ these materials even without a rigorous theoretical understanding – which was the situation prevailing from the invention of the first Al-Cu alloys, as the still-familiar 'Dural', first used in the Graf Zeppelin, which soon found universal use in aircraft after the very earliest days of flight.

It is vital, however, to appreciate the limitations imposed with joining by welding. The highest-strength common alloys (such as 2014 and 7075) provide an enormous strength-to-weight benefit, being one-third the weight of a C-Mn steel and possessing comparable strength, but they cannot be welded. Hence, for aircraft, joining by bolting, riveting and adhesives is still widely used and can be a perfectly satisfactory solution to the fabrication problem. Where it is preferable to produce butt joints (as in cycle frames) then only the commercial purity (CP) 1000-series or the 5000-series (e.g. 5083) are preferred, although some others (like the 6000) are weldable but only with significant strength loss, as-welded, and require a difficult heat treatment to restore properties to that of the strengthened parent metal. The 2000- and 7000-series are *unsuitable* for welding. The best choice for full-strength joints is the non-heat treatable 5083, using filler wire such as grade 5356. This can provide a tensile strength of ~45 ksi (300 N/mm²) – well over twice that of the weldable heat-treatable alloys (like 6061) where strength is reduced to the ‘O’ temper, unless heat-treated after welding – often not a practical proposition, with potential problems of distortion.

Magnesium (Mg) alloys

Magnesium alloys present the greatest potential for the highest strength-to-weight ratios of the metals we are likely to come across, and has found applications in critical parts, from aircraft under-carriages to racing cars and motorcycles of surprising vintage, such as the iconic pre-war Manx Nortons and KTT Velocettes. The equilibrium diagrams show similarities to the wrought Al alloys, and a similar precipitation hardening mechanism is employed to achieve the remarkable strengths, approaching 60 ksi/400 N/mm², with 10 per cent elongation, of the best alloys. And this is combined with a weight of only 62 per cent of the aluminium alloys.

Despite some highly attractive characteristics these alloys have failed to gain the widespread use of the aluminums, partly due to

difficulties with self ignition during machining unless particular precautions are employed. Small particles can be prone to combustion, but normal pieces are entirely safe in this respect; some problems with corrosion that require careful evaluation have not helped with wider application.

This should, by rights, be a most common material but in practice remains something of an enigma for the non-specialist. Whilst the above-mentioned examples demonstrate that it can be an optimum choice for critical weight-saving problems, this is presently limited to castings and pre-formed wrought items.

Considerable research is currently under way to extend application in sheet materials, particularly important for volume-production motor vehicle body panels. The problem arises from the hcp (hexagonal close-packed) atomic structure, which allows slip (deformation) primarily on the basal planes, making the fabrication of sheet material difficult, as the basal planes in sheet are aligned approximately parallel to the sheet axis (see [Chapter 5](#)). Basically, this means that deformation is difficult to control, as properties vary according to the microstructural orientation, unlike comparable materials like steel and aluminium. Current research concentrates on introducing discrete particles or alloying with 'rare-earth' elements, and thus is a long way from ready availability.

Titanium (Ti) and Ti alloys

Although a somewhat higher-density element than the preceding light metals, Ti has some highly desirable properties that can more than make up for this deficit, primarily as one of the most corrosion-resistant of the metals, combined with high strength and good workability. Even joining is a practical possibility; once the dramatic affinity of the molten metal for O₂ contamination is understood, the 'myth' of difficult welding can be overcome, with considerable care in gas-shielding *all* around, and TIG welding. The greatest drawback in its employment for a very wide range of applications is simply its indisputably high cost.

Titanium is perfectly usable in its (virtually) pure form, with only small additions of O_2 . These provide a range of strengths – Ti-grade 1, 2 and 3 – with increasing O_2 providing strengths from ~50–70 ksi (300–500 N/mm²) combined with good ductility. These are so stable (corrosion-resistant) that they are often found in critical body implants, such as bone-plates or hip-replacements. Industrial applications include severe marine and petro-chemical environments, with highly stressed heat-exchanger internals utilizing the combination of strength, lightness, corrosion resistance and weldability.

The structure of the Ti- O_2 alloys is a simple alpha (α) phase up to 880°C, which is reflected in its ductile characteristics. Where increased strength and severe service is required, encountered in aircraft structures and the front sections of jet turbines, a series of Ti alloys has been developed, such as the Ti-6Al-4V ('318') $\alpha + \beta$ alloys. In these alloys precipitation-hardening provides extremely high strengths of over 200 ksi (1500 N/mm²). Although these have more limited ductility, it is not a problem as they are invariably formed by hot working (see [Chapter 5](#)).

Nickel (Ni)

Nickel in its pure form has similarities with mild steel – comparable melting point, ductility, work-hardening characteristics – with a slightly higher density (8.9 g/cm³), but lower magnetic permeability. Its resistance to corrosion, however, is the key to its value. When used as an electroplated layer it provides a high level of resistance to atmospheric corrosion and was extensively used in plating steel for early vehicles. Indeed, many an apparently authentic restoration of a vintage vehicle has been exposed by the bright Cr plating that has become the near-universal method since the 1930s.

Nickel alloys

The major practical use of Ni is when alloyed, firstly in small proportions, for low alloy steels and cast irons, conferring improved

hardenability and toughness; and secondly, the family of Ni alloys with a high percentage of Ni (often >50 per cent), where the addition of strengthening elements such as Cr, Mo and Co have led to the development of a critically important series of alloys that can operate at the extreme temperatures required in the modern jet engine. The corrosion resistance and high temperature ('creep') strength has led to extensive use for turbine blades. These still retain their preeminent position as the material of choice for the most arduous conditions, retaining sufficient toughness to withstand the impact tests that have hindered several promising alternative high temperature materials, like ceramics and fibre-reinforced metal-matrix composites. The failure of promising fibre-reinforced compressor turbine blades, due to lack of toughness during bird-strike tests, almost led to the demise of the great aero-engine maker Rolls Royce in the 1970s, now in the forefront of commercial jet engine manufacture. Current Ni alloy turbine blades can operate at around 1600°C, some 150°C above melting point, utilizing internal cooling passages enabled by highly developed investment casting techniques. Whilst this development of exotic alloys is not of immediate import to the home engineer, primarily because of the high cost of these special materials, some alloys have been sufficiently widely used to make it possible to find product that can be recycled for our purposes. Such alloys could include UNS NO6625, as Inconel 625, and even the Hastelloys, like UNS NO10276, as Hastelloy C276.

Table 14 Nickel.

Type	Composition	UTS (N/mm ²)	Elongation (%)
Nickel (commercial purity)	99.5Ni, Si, Mn, Fe	345	30
Monel (typical)	65 Ni, 32 Cu, Si, Mn, Fe	1000	20
Constantan II	50 Ni, 50 Cu	585	25
Nichrome (typical)	80 Ni, 20 Cr		
Haselloy (typical)	65 Ni, 20 Mo, 15 Fe, Cr, Cu, W	1200	15
Inconel (typical)	45 Ni, 25 Co, 25 Cr, 10 Fe, Ti	1250	30
Nimonic (typical)	72 Ni, 18 Cr, 2 Ti, Fe, Nb	1100	30
Invar (typical)	40 Ni, 50 Fe, Co, W, Cr		

It is useful to bear in mind that material selection for severe applications often owes more than might be imagined to ‘fashion’, albeit not intended pejoratively. Exotic material selection may not be as quixotic as those of the catwalk, but nevertheless there is a degree of ‘following the herd’. This may, not unreasonably, lead to extensive use of, say, alloy 625 for many severe marine applications; C276 for petro-chemical offshore service; and even ‘super duplex’, like Zeron 100, for petro-chemical pipelines. It should not be imagined that these, necessarily, represent the only choice for an application, and so the home engineer should not be too obsessive in material selection – even for perhaps our nearest equivalent, the model jet engine. Many of these high Ni alloys could be pressed into service bearing in mind the old motto that ultimately, there is ‘no substitute for experience’, even with modelling or testing!

Monels, Nichromes, Inconels and Hastelloys

Monels comprise Cu-Ni alloys and, as might be anticipated, exhibit excellent resistance to corrosion in marine environments, and food handling. Used for handling hydrochloric and sulphuric acids, even de-aerated hydrofluoric acid, they find widespread use in the chemical industry.

Nichromes are Ni-Cr alloys which exhibit high electrical resistance, combined with resistance to oxidation, used for electrical heating elements and resistances.

Nimonic and Inconels, as noted above, can retain strength and toughness at exceptionally high temperatures. This is particularly important with regard to providing creep resistance, where the material elongates during prolonged exposure to stress at high temperature, the scenario for a jet-aircraft turbine blade as above, with ratings based on 10,000 hour testing. Whilst creep is not normally considered a characteristic encountered by the non-specialist, its effects can be readily seen in, for instance, historic lead domestic piping which can droop between supports; or the tungsten (W) filaments of (non-energy saving!) light bulbs. It has even been cited as a major contributor to the 9/11 collapse of the twin towers in New York.

Hastelloys are the series of Ni base alloys developed by the Haynes Corporation, Huntingdon, USA; they are similar to the Inconels and are widely used in the most severe applications of the petro-chemical industry, especially oilfield exploration. Their high alloy (Ni, Cr, Mo, Co, W, etc.) content make these inevitably expensive and their high strength does not lend them to machining in anything other than the most exotic home workshop!

Lead, tin, zinc and bearing alloys

Lead (Pb)

The use of lead dates at least from Roman times, with many examples of pipework and artefacts excavated from ancient Rome, Pompeii, and throughout the Roman Empire, almost exclusively found in pristine condition, as evidence of its remarkable corrosion resistance in soils. Subsequently, roofing materials from the medieval period have continued to perform well into the modern era. So well, in fact, that the recent blight of metal-stealing from church roofs has been the greatest cause of failure! A problem of creep, as

noted earlier, is perhaps one of the greatest drawbacks when used in such applications.

Commercial lead, usually produced from the mineral galena (PbS), is noteworthy for its extremely high density, low melting point, and corrosion resistance, combined with low strength and high ductility. It is widely used for protection from radiation in radiography, for industrial and medical use, and for handling the highly corrosive phosphoric, chromic, sulphuric and sulphurous acids.

The low strength of Pb enables easy working, and it is almost unique in having such a low re-crystallization temperature that its stress–strain curve shows a virtually flat plastic-region, indicating virtually no work-hardening. This makes it a valuable material for studying fundamental metal-processing methods such as rolling. A billet of Pb behaves, at RT (room temperature), in the same manner as steel at red heat (~800°C). Pb is readily available, at least in rolls of sheeting materials, from plumbers' merchants.

Table 15 Lead.

Type	Composition	Applications
CP Lead	99.99 Pb	chemical industry
Arsenical Lead	99.6Pb, 0.15 As, 0.1 Sn, 0.1 Bi, Cu, Te	cabling, improved strength CP
Antimonial Lead	90 – 99 Pb, 1 – 10 Sb	batteries, cables, extrusions
Solder (traditional)	60 – 90 Pb, 10 – 40 Sn	joining
Babbitt	60 – 80 Pb, 10 – 15 Sb, 1 – 10 Sn, 1 – 3 As	bearings
Type (printing)	70 – 90 Pb, 10 – 20 Sb, 2 – 10 Sn, 0 – 2 Cu	traditional printing industry, improved strength

Woods Metal (traditional)	20 – 40 Pb, 30 – 50 Bi, 10 – 20 Sn, 0 – 30 Cd	low melting point alloy
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Tin

While tin is an important element, its cost and low strength finds use primarily as a coating of steel for corrosion resistance (such as food cans); alloying with Pb for solders; as well as in just about all the alloying systems described so far. Pure Sn can undergo an ‘allotropic’ transformation at 13°C, from its body-centred tetragonal structure to a brittle cubic alpha phase (see [Chapter 4](#)). This is avoided by small additions of bismuth and antimony, making it highly stable, as shown in cans of food still pristine after a hundred years (though presumably past their sell by date...).

Another important use is in bearings when combined with copper and antimony.

Table 16 Tin.

Type	Composition	Applications
CP Tin	99.98 Sn	high corrosion resistance, tin foil, plating
Babbitts Tin	60 – 90 Sn, 5-15 Sb, 3 – 5 Cu, 0.15 As, 0.1 Sn, 0.1 Bi, Cu, Te	bearings
Solders	60 – 90 Pb, 10 – 40 Sn	joining
Die Cast alloy	82 Sn, 13 Sb, 5 Cu	die casting material
Pewter	91 Sn, 7 Sb, 2 Cu	decorative work, non-varnishing, non-toxic

Zinc

Another white metal, like Sn, Zn is a heavy element that finds considerable use as a casting alloy where its low melting point, relatively high strength, and corrosion resistance, make it easy to cast intricate items like the carburettor bodies (as in [Case Study 3, Chapter 1](#)). These are almost invariably produced by die casting.

The use of Zn plating on steel provides a superb method for protection of the steel substrate, and this ‘galvanizing’ treatment has been largely responsible for the step-change in the car industry, from the rust-buckets of the 1950s–1970s, following the universal adoption of monocoque design. Such car body corrosion is, thankfully, now virtually extinct – despite use of higher strength, thinner, sections.

The many other uses include manufacture of dry-cell batteries, and as a major alloying element for the brasses.

Table 17 Zinc.

Type	Zn	Al	Pb	Applications
CP Zinc	99.9		0.1	drawn product, sleeves
AG 40	96	4		vehicle parts, general hardware
AG 41	95	5		die casting material
Casting alloy	94.5	5.5		slush castings, domestic products

Bearing materials

This is the final section of the common materials we are considering, one that is vital for the function of many machines and which spans a remarkably wide range (from the high strength steels typical of rolling bearings, as ball- and roller-races, to the soft-metal, plain bearings – and in between a variety of bushes, often from Cu-base alloys. These are found whenever there is a rotating or reciprocating shaft – which is, by definition, in every machine.

The rolling bearings can be considered a special case, as fundamentally they rely on rolling elements and are typically characterized by very hard and fatigue-resistant low-alloy steels for their rolling elements and races. We consider them under their steel analysis in [Chapter 4](#). By contrast the ‘bushes’ and soft metal plain bearings achieve their purpose by presenting a large surface to the rotating, or reciprocating shaft.

The characteristics of plain bearings include: surfaces that will allow sliding without pickup (‘galling’, wear caused by adhesion between sliding surfaces, is common when stainless steels are brought together under load), undue friction, deformation, or wear. Simple bushes are often made from one of the Cu-base alloys, when supporting hardened steel shafts.

For more extreme loading, as in the crankpin end of the connecting rods of an internal combustion engine, use can be made of the specially developed ‘shell bearings’. These comprise a strong steel shell as a substrate onto which is deposited a carefully selected soft white-metal facing, such as an aluminium-base Al-Sn alloy, or copper base Cu-Sn-Pb alloy. This presents a large surface area, thereby reducing local stress on the crankpin, in contrast to the roller bearing design. The soft metal presents a low-friction sliding surface providing that sufficient lubrication is maintained. The entire success of plain bearings and bushes is critically dependent on the role of lubrication. In practice, this has been proven so effective that extreme loading conditions can be accommodated to provide engine lives of hundreds of thousands of miles – at a wear rate approaching zero!

Table 18 Bearing materials.

Type	Composition
Tin babbitt	60 – 90 Sn, 5-15 Sb, 3 – 5 Cu, 0.15 As, 0.1 Sn, 0.1 Bi, Cu, Te
Lead babbitt	60 – 80 Pb, 10 – 15 Sb, 1 – 10 Sn, 1 – 3 As

Tin bronzes	85 – 90 Cu, 10 – 13 Sn, 1 – 3 Zn, 0.1 – 1.0 P
Copper-lead alloys	60 – 80 Cu, 20 – 40 Pb, 0.5 – 2.0 Ag
Aluminium alloys	90 – 92 Al, 5 – 7 Sn, 0.7 – 1.3 Cu, 0.5 – 1.5 Ni, Si, Mg
Impregnated bearings	90 Cu, 10 Si + graphite (sintered bronze); 96 Fe + graphite, etc (sintered iron)

White metal alloys

These are often known as ‘Babbitt metals’ after Isaac Babbitt, who filed the original patents for these materials, later developed by Vandervell Industries into disposable shell bearings; the proceeds from this highly successful industry assisted the development of the iconic ‘Vanwall’, the first Formula One championship-winning manufacturer from the UK, in 1958. The bearings have a Cu-, Sn- or Pb-base soft matrix, with embedded hard crystals such as antimony (Sb). These hard particles provide strength, rather like stones in concrete, whilst the soft matrix conforms to the profile of the shaft. In addition, any particles entrained in the lubricant can become safely embedded in the soft matrix, thus minimizing scouring of the shaft.

Self-lubricating bearings

Often referred to as ‘Oilite’, these are made by the ‘sintering’ process, where the compacted powders, normally Cu-base alloys, are processed to part fully-sintered state whilst leaving deliberate porosity of around 30 per cent. The porous structure is then impregnated with a lubricant, retained in the voids. This leaches out in service, thereby providing enough lubrication for applications where no other lubrication system is provided.

Special cases

There are, by contrast, many applications of sliding surfaces where friction is deliberately utilized, such as brake drums where grey cast iron is universally employed. The free-graphite provides sufficient means to avoid adhesion between the friction material and drum or disc, whilst still providing a friction force that generates the braking effect. This same property of limited exposure of the graphite flakes also makes grey cast iron the material of choice where metal-to-metal friction needs to be reduced. When combined with sparse lubrication, in the arduous applications of cylinders with pistons that contain pressure with high sliding speeds and loads, it enables operation without adhesion and seizure. The typical internal combustion engine aluminium piston accommodates grey cast-iron 'compression rings' that provide the sealing element, and yet permit sliding in the cast iron cylinder or liner.

These cylinders can be formed as thin sleeves or liners, pressed into light-alloy (typically Al-Si) castings. They may be heavily finned (for air-cooled engines) or used as a water-cooled aluminium 'cylinder block'. This provides the best contact surface, combining the weight-saving advantage with enhanced thermal conductivity of the aluminium cast cylinder or block.

Finally, very hard materials such as sintered carbide or diamond can be used for specialist thrust bearings (familiar in pre-digital watches), where high loads and limited lubrication are the norm. This is, clearly, a very specialist area and not usually the province of the home engineer.

METAL FORMING PROCESSES

The properties and application of materials is largely dependent on the method of manufacturing. Product can only be realized, for practical purposes, by one of the fundamental processes of either casting or 'deformation processes'. Deformation processes include forging, rolling, extrusion, cutting, machining, sintering (compaction

and heating of powders to produce solids), and some special processes of less import here, such as high- or low-energy explosive or vacuum-forming. Recent developments like the much-hyped ‘additive manufacturing’, or 3-D printing, whilst they are interesting cutting-edge technology, will probably not take long to reach the more well-heeled shed. The associated processes of joining, such as welding, are examined in [Chapter 5](#).

Casting

This is a truly ancient process, although the very earliest metals, around 10,000BC, were simply ores subjected to hammering with stones, or limited heating. The first castings were produced c. 5,000–3,000BC, before the Bronze Age.

The essence of casting is to produce a fully fused melt of one or more elements which is then transferred, usually by pouring into a mould, of which many types exist. This is followed by solidification, and finally removal from the mould, by destructive or other means, to produce the ‘as-cast’ product. This may, or may not, according to the process and product requirements, produce the finished item, as-cast.

There are some fundamentals common to this diverse range of casting processes, the first being the solidification characteristics, with regard to final size. Almost all metals undergo a volume-change during solidification – usually, but not always, contraction. This needs to be taken into account, otherwise the final size may be an unpleasant surprise – discovered when insufficient metal is available for machining a sealing face! Standard practice is to use ‘contraction rulers’, enabling the pattern to be made to a corrected size, resulting in the casting matching drawing dimensions.

Another feature of casting is the metallurgical effect of the solidification process, especially where the item is used ‘as-is’, and subjected to no further significant processing like heat-treatment or forming. Some fundamentals of the solidification process are described in [Chapter 5](#), but in this general overview we can note that

the different cooling rates arising from the item shape, making use of the forming complexity unique to this process, can lead to pronounced 'inhomogeneity' (microstructural variability). Generally, large and simple sections cool at the slowest rates whilst thinner sections, naturally, solidify earlier and cool faster in the solid state. The structure produced is dependent on the alloy system; for instance, aluminium-silicon alloys benefit from rapid cooling, and chilling blocks are often used to control the direction of solidification and avoid shrinkage cavities. Other materials, notably cast iron, can produce extremely hard structures when too rapidly cooled, due to the high carbon content.

A particular feature of casting is the preference for solidification to propagate as 'dendritic growth', in a 'fir tree' pattern, with solidification proceeding along and across branches, as these shoot out into the liquid, from nucleation sites or the near-surface of the mould (see also [Chapter 5](#)). The general effect in castings is for relatively large grain sizes and tendency for segregation of lower melting point phases to be trapped in this growing dendritic structure, as solidification advances towards the centre of the melt.

All of these effects tend to make for inferior mechanical properties in castings, compared to 'wrought' product – though all products, apart from 'sintered', originate from a cast, of one sort or another. Wrought items derive from castings, 'ingots', formed by heat and pressure, thus manipulating the microstructure into a more favourable state. Judicious working and heat treatment through 'recrystallization', can provide an 'equiaxed' (randomly oriented) and homogenous grain structure.

Despite these potential pitfalls, such product can satisfy its design requirements as-cast, and considerable expertise has been developed in foundrywork over the past 150 years or so. An example of a supremely efficient casting is the fusion weld – the weldment being simply a specialized form of casting, effectively as-produced from a miniature arc furnace. Mechanical testing demonstrates that the weld, provided it is properly chosen, is normally superior to the base metal.

This is, admittedly, an example of a very small casting, and evidence abounds for ultra-high performance items such as the single crystal jet turbine blade, but as casting size increases, so do the potential difficulties in producing a high integrity product. This can be perfectly acceptable when as-cast is the only practical solution. Fabrication from an assembly of smaller parts may often be difficult or even impossible – as for ships' propellers or church bells.

Sometimes it is only possible to get a consistent microstructure, with the required phase composition, by heat treatment. This can apply to the stainless steels, which can precipitate undesirable ('deleterious') phases during cooling, requiring water-quenching from $\sim 1,100^{\circ}\text{C}$. However, many of the alloys with which we are concerned, such as aluminium-silicon and some copper-base alloys, can be used in the as-cast condition. Note that these tend to be items not subjected to particularly high stresses – such as cylinder heads for two-stroke internal combustion engines – easily cast but impractical or impossible to fabricate or machine from solid. (See [Chapter 1, Case Study 3](#).)

Cast iron, such as grey cast iron, is a good example of a material that can be produced relatively simply in large and complex shapes, provided that its inherent lack of ductility and toughness is recognized. It provides the ideal material for parts subjected to heavy compressive stress, such as machine-tool beds. The graphite flakes, responsible for the poor toughness, provide a positive benefit in accommodating vibration loads, a high 'damping capacity', and self lubrication, for metal-to-metal sliding.

The variability of grey cast iron, as a low-cost material, is a consequence of specifications, as seen in UK or USA national standards, which simply define grades from a range of tensile strengths, with no limitations on compositions or foundry process. This potential variability in hardness can make machining problematic for the home workshop with only a small lathe on which to turn a large item! Similarly, the welding of cast iron (of which more in [Chapter 5](#)) is normally tricky at best – not least because of the intrinsic problems with extreme hardenability in the HAZ due to the

high carbon content of 2–4 per cent. This composition variability makes it a task only for the experienced, bold, or lucky!

Sand casting

This is the oldest casting method, employed since antiquity, but still perhaps the most widely used in commercial foundries worldwide. In its traditional form it utilizes a pattern that is impressed into a split casting-box filled with sand. The shape left by the pattern when removed from the split casting box, plus ‘runners’ and risers’ (entry and exit passages for the liquid melt) is then ready, on re-assembly of the casting-box, to receive the molten metal, poured from the ladle of melt from the furnace-heated crucible.

Success is largely governed by the selection of an appropriate sand that, when mixed with a binder, allows retention of the shape of the pattern when removed from the casting-box, leaving the void into which the metal is poured. This sand mould must accommodate the molten charge without undue reaction, whilst being sufficiently porous to enable transfer of gases without entrapment, and thus porosity, in the casting.

Pattern-making requires considerable skill and experience, in order to provide a profile that will accept the fluid without interruption to mould filling. The design should avoid re-entrant shapes and too-rapid changes of section that lead to seriously unequal cooling rates. Fluid flow is provided by careful design of runners and risers, with foundries now commonly employing well-proven computer modelling to design these feeders. The riser provides a reservoir to accommodate local shrinkage during solidification, leaving the casting free of shrinkage voids.

A further feature is the use of ‘cores’, usually prepared from shaped sand, often as bars, which are introduced (carefully!) into the casting box after removal of the pattern and before closure and clamping, ready to receive the melt. The cores are then enveloped by the fluid and later removed by breaking out, after the casting has been broken out from the mould. Complex internal shapes can thus

be produced to finished size, avoiding difficult or impractical machining, as commonly encountered in ports and water jacket passages in internal combustion engines.

Sand casting is therefore a highly adaptable and flexible process, well suited to small-volume production. The pattern is often made from wood, but harder-wearing materials are preferred for extensive re-use. In the future, the prospect of using additive machining has significant potential for pattern making, but this is some way off for the lone operator. Whilst potentially a viable process for the home workshop, melting and mould pouring is not a natural activity for the machining-orientated home engineer, so finding a friendly sand-cast foundry is a valuable asset.

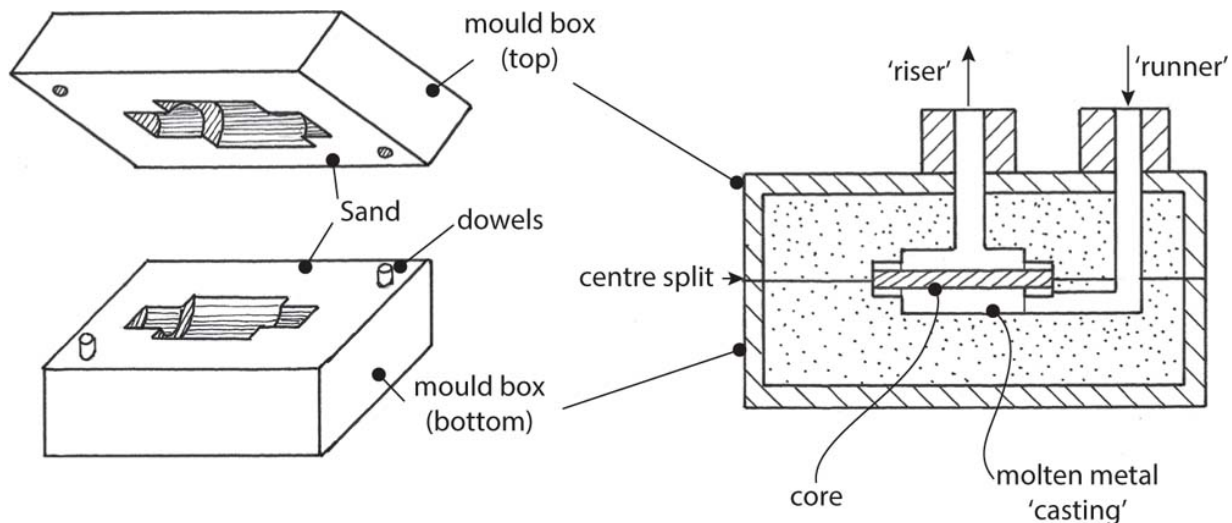


Figure 2 Sand casting – principal elements.

Investment casting

For a process in use since at least 2,000BC in central Europe, the Middle East and China, it is a remarkably sophisticated means of producing stunningly detailed and accurate, albeit usually small, castings.

Compared to sand casting it offers a fine surface finish and accurate replication of detail that has long made it a prized method of producing high-value jewellery and domestic artefacts. In modern

times, it has provided the method for a wide variety of complex industrial castings, including the iconic turbine blades mentioned earlier. Although an intrinsically low-volume production method, nevertheless it has been adapted to accommodate commercial batch volumes by coupling a series of individual moulds to a common-feeder system. And even in ancient times, in Thailand and China, large ritual statues were being cast in copper-base alloys – typically in large, thin sections, of great sophistication – joined by soldering into the huge religious effigies that have survived to modern times.

Investment casting uses a pattern in the same principle as sand casting, to provide the impression of the product, in a casting box. However, the pattern is made from a wax-based material which is produced either by carving from the solid or from injecting the wax into a die, as often used when more than one-off reproducible products are required. The wax pattern is then ‘invested’ – a process involving dipping into a series of slurries which essentially comprise ceramic particles suspended in water. The dipping and drying process is repeated according to the thickness of investment required – it is this ‘shell’ that provides the mould into which the molten metal is poured.

Removal of the wax, thereby leaving the investment shell, is achieved by heating the mould and allowing the wax to run out of the feeder arms. The surface finish and detail can therefore be finely controlled, from the characteristics of the first investment layer. The process also enables passages to be produced by building the wax pattern over ceramic cores. As an example of a most sophisticated product, the modern turbine blade uses a complex structure of these extremely fine cores to provide passages in the finished casting that permits the flow of cooling gas through the elegant shape of the blade, thus enabling the critical cooling, and higher operating temperatures, that contribute to the vital efficiency improvements in jet engine performance. The wax is injected into a mould that replicates the blade shape, and the solidified wax is then ‘invested’ as described. For the most critical, highest temperature blades, the nickel-based alloys are poured into the investment shell held in a

furnace where a carefully controlled temperature gradient and 'seeding' is employed to produce a casting in its finished state as a single crystal – the optimum structure for creep resistance.

Die casting (permanent mould)

In die, permanent-mould casting, the cavity into which the molten metal is poured is cut directly into the mould. Instead of the sand, therefore, the mould comprises a split metal die. Therefore, no pattern is required and the split die, comprising two or more parts, can be rapidly assembled as it contains all the accoutrements of the casting process – runner, mould cavity and riser, in a permanent die arrangement. Accurate alignment of the die sections is provided by dowels and the assembly held together with clamps. The complete facility is re-usable and adaptable to a variety of sizes, principally limited by the cost and difficulty in producing the die in a material which will resist the molten metal, clearly a difficult and expensive business. Cast iron, steel and graphite are commonly used for the die. These materials are employed for dies to cast the lower melting-point alloys, including aluminium, magnesium, and the zinc-base alloys familiar in complex, but low-stressed, carburettors for internal combustion engines, as [Chapter 1, Case Study 3](#).

Die castings are notable for their high dimensional accuracy and surface finish. These attributes reduce machining times for cleaning up sealing faces and can provide as-cast product of fine surface finish. The potential for a high volume production rate also make it attractive in white goods and vehicle manufacture, but it is not the obvious choice for the home workshop, due to the time and complexity of die manufacture. However, it may still be attractive for relatively small batch manufacturing of specialist items such as the model aero engine shown in [Photo 26](#). The time invested in producing the small die is rewarded by the fine finish, accurate dimensions and appearance appropriate to a precision product demanding high efficiency in manufacture to compete in a cost-sensitive market.

The die structure also lends itself to either gravity, as with the previous methods, or forced injection, of the liquid metal, typically at pressures between 5 and 20 N/mm² and in section thickness from <1mm to >10mm. A further useful facility is the use of a vacuum to draw in the fluid, especially for small, intricate products. The use of pressure or vacuum promotes high integrity of mould filling and consistency of casting lending itself to automated processing. A particular feature is the intrinsically high cooling rate in this method, although preheating of dies can be used to mitigate this effect. (See [Bibliography](#) for practical processing.)

Further casting methods

There are many variants of the above processes, often developed for particular products, such as the high-volume production required for large castings of great complexity and high integrity. This includes internal combustion engine cylinder blocks for cars, now often involving special 'squeeze' processes that combine elements of casting and forging.

Centrifugal and 'continuous' casting are used to produce castings with particular properties or extensive lengths of product, but do not bear directly on our applications. By contrast, the 'lost foam' process is very well suited to one-off manufacture at minimum set-up cost. This process uses expanded polymeric material, such as polystyrene, as the pattern. This can be moulded by hot-wire cutting, or carving with knife or saw, and glued together to form the required general shape, although it certainly doesn't expect to provide the detail replication of the 'lost wax' process. The pattern is treated similarly to that of sand casting. A simple and effective process, appropriate to an item that is required reasonably near finished size, it can provide a possible solution to some awkward tasks encountered by the home engineer (see [Bibliography](#)).

Deformation processes

Forging, rolling, extrusion and drawing

These are the classic deformation processes, employed to change the raw material (ingot) into the required shape, or sufficiently near to facilitate finishing operations (such as machining or hand-working). The first requisite is sufficient ductility of the material, predominantly in compression mode, to accommodate the deformation without fracture.

The effort required to deform the metal governs the methods and machinery of the selected process. As almost all metals work-harden to a greater or lesser extent, this makes extensive or repeated operations increasingly difficult; as well as moving the microstructure nearer to its ultimate strength limit. Therefore, it is common to apply significant heating, to around $0.5 \times$ melting point (T_m), which serves to reduce the yield point and can also play an important part in effecting desirable changes in the microstructure during the final operation. This latter effect is shown in [Chapter 5](#), under 'Re-crystallization', where the mechanism of limited deformation and heating can produce a greatly improved microstructure, rendering a product superior after rolling by virtue of controlled grain refinement with improved mechanical properties – magic!

The use of heat to simply overcome the effects of work-hardening, as well as reducing the yield stress to permit working at lower loads, has been understood since antiquity and is the cornerstone of the blacksmith's art and craft. In essence, present-day forging processes are a mechanized extension of blacksmithing, enabling heavy deformations without corrupting the material's structure – provided that working is undertaken at the appropriate temperature. Whilst this is around 'cherry red' ($\sim 800^\circ\text{C}$) in many steels, other alloy systems (such as copper-base alloys) require a detailed knowledge of the equilibrium diagram to ensure that working in the required phase is practised (see [Chapter 5](#)).

In *forging*, the blacksmith's hammer is replaced by a hydraulic ram (that may contain a die), to work the material to the desired shape,

such as dishing a flat plate for pressure vessels, to sizes simply limited by the power of the press. In common with the other deformation processes, apart from recrystallization there is the benefit from the flow pattern of the microstructure. This can be seen by cutting through the section, polishing and etching, as the macro-structure is often visible without magnification.

Rolling involves the compression of the ingot between two powered rollers, to produce a uniformly reduced thickness. Virtually 100 per cent of the deformation is accommodated with extension of *length*, uni-axial deformation, whilst retaining the original width. Repeated passes may be made to reduce the thickness from the original ingot to thin sheet, provided appropriate inter-pass heating is employed. Well-established techniques enable hot- and cold-working to be used for steel plate and sheet to provide a wide variety of properties to designer requirements, as in 'TMT', thermo-mechanical treatment, rolling.

Extrusion is used to force the ingot through a die, whereupon it assumes a cross-section as provided by the die aperture. Where simple section-shapes are required, such as round or square bars, the rolling process is adapted with rolls replicating the section required. This can then be developed in small sections by 'drawing' the bar through a die, enabling successive reductions to form very high strengths of piano wire, noted earlier.

Another version is *deep drawing*, where flat material is pressed through a die, typically to make metal cans, as in [Figure 3](#). This method takes a flat sheet and forms it in a series of operations into the finished shape, by careful control of the punch, die and pressure pad. This process is used extensively for producing thin-walled cans for food and drinks where multiple operations provide high production rates – although not usually a method that finds direct application for the home engineer.

The characteristics of the item after these deformation processes can therefore be dramatically changed from the original ingot, and requires an understanding of the response of the material to deformation at different temperatures – the work-hardening at lower

temperatures being attributable to dislocation activity, as described in [Chapter 5](#). Heat treatment after deformation is also addressed in [Chapter 5](#), so that properties can be tailored to the application. For instance, a cold-rolled steel sheet, familiar in white goods or car body panels, will distort dramatically with local heating, e.g. welding, partly due to release of the residual stresses imparted by the final, cold-rolled and/or pressing operations.

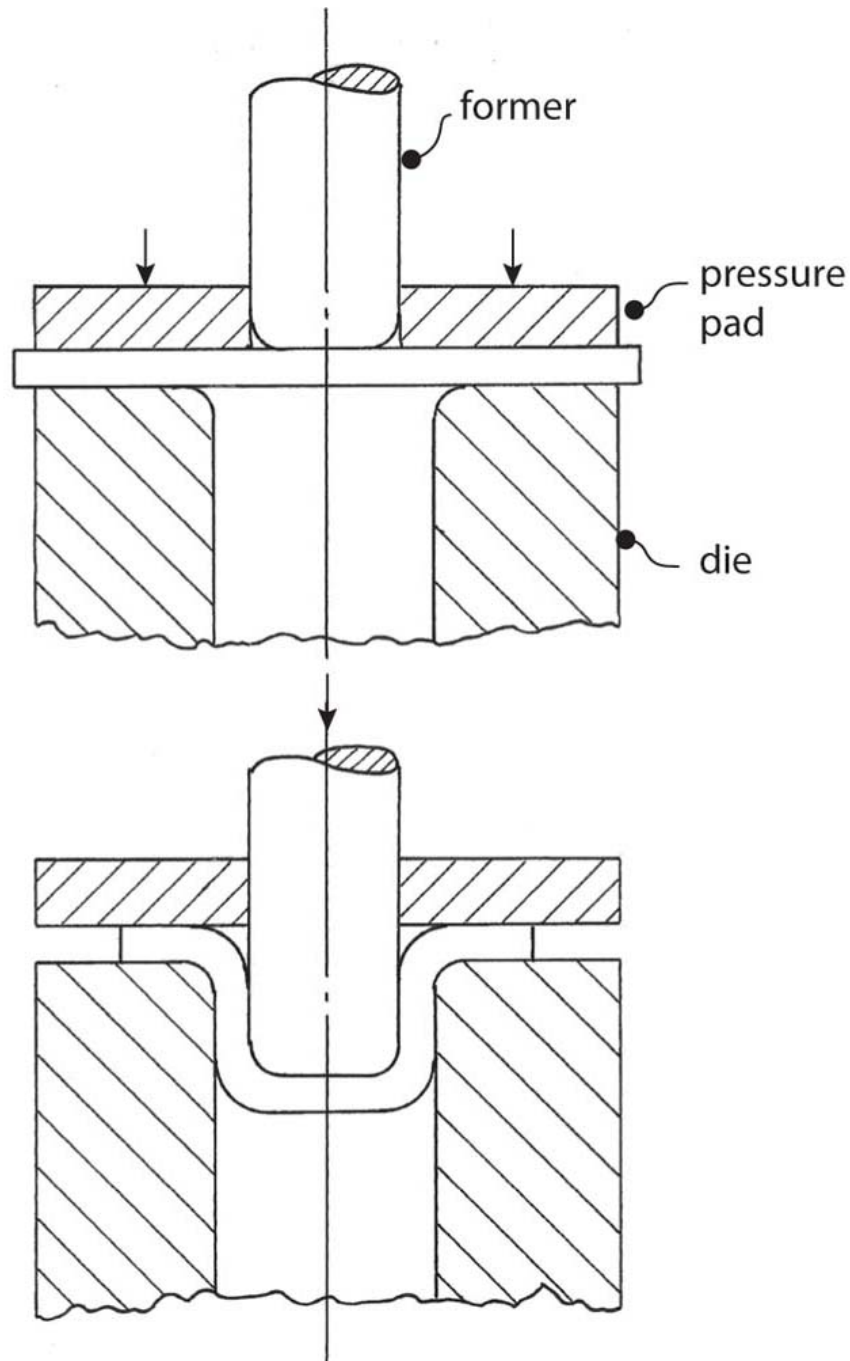


Figure 3 Deep drawing principle.

Cutting and machining

While the home engineer is usually very familiar with the usual cutting and machining processes (sawing, filing, drilling, turning, milling and grinding), and is well catered-for in the literature with

regard to the enormous variety of tooling and process details, nevertheless we can offer some fundamentals from the *material's* viewpoint.

Unlike the preceding processes where deformation produces no material volume change, rather a change from one shape to another, cutting and machining are essentially concerned with 'shearing' processes in the material. The former processes normally impact on the material's properties, to such an extent that heating during processing, and heat-treatment can be a vital factor. In cutting and machining, however, whilst there is some minor deformation, the cutting element – be it a punch, saw tooth, lathe tool tip, or grinding wheel grit – is designed to slice through a part of the material. Therefore the effect on material properties can often be neglected.

An important attribute of the material is therefore a measure of its machinability, and a grasp of this property is inseparable from the task of shaping metals by these means. Machinability is generally promoted by low strength and moderate ductility. The preferred behaviour, in a lathe tool, for instance is: firstly, a low cutting-force, less demanding of machine power and rigidity; and secondly, discontinuous 'chips'. The more ductile material will tend to spread over the cutting face, requiring more energy, motor power, to progress the cut. The material strength governs the shear force required to provide separation of cut material from the base metal and initiation of the chip. A very soft, ductile, material will therefore have a lower machinability index than a less ductile one; and second-phases can be useful in providing a microstructure with discontinuities acting as chip-breakers, to avoid the swarf trails that are prone to tangling around the cutting tool. Clearly, higher-strength materials are progressively less machinable with increasing strength, as typified by the high-strength stainless steels and nickel-base 'superalloys', which combine both high strength and toughness, making them difficult to machine without powerful and expensive machines and specialized tooling.

Even within a single alloy system, machinability can vary dramatically, with an obvious example being cast iron. Even grey cast iron can range from poor machinability at its highest-strength specification, despite low ductility, to good machinability for a lower-strength version. Where cooling rates of cast iron are high, during casting thin sections or as a result of welding, then White cast iron may be formed. This then becomes effectively un-machinable, without resort to grinding, which usually implies low material removal rates.

In order to achieve the free-cutting versions of plain carbon and alloy steels, it is common to use small additions of insoluble elements, like lead or sulphur in ENIA and 303 stainless steel. This provides a material that is a pleasure to machine, but renders it unsuitable for welding and may cause problems with heat treatment, so care must be exercised in their use.

Table 19 An approximate guide to machinability.

Material	Machinability index
Stainless steel	40
Wrought iron	50
Copper (part worked)	60
Aluminium bronze	65
Cast steel	70
Mild steel (free cutting)	100
Brasses (free cutting)	200

(The higher the number, the better machinability.)

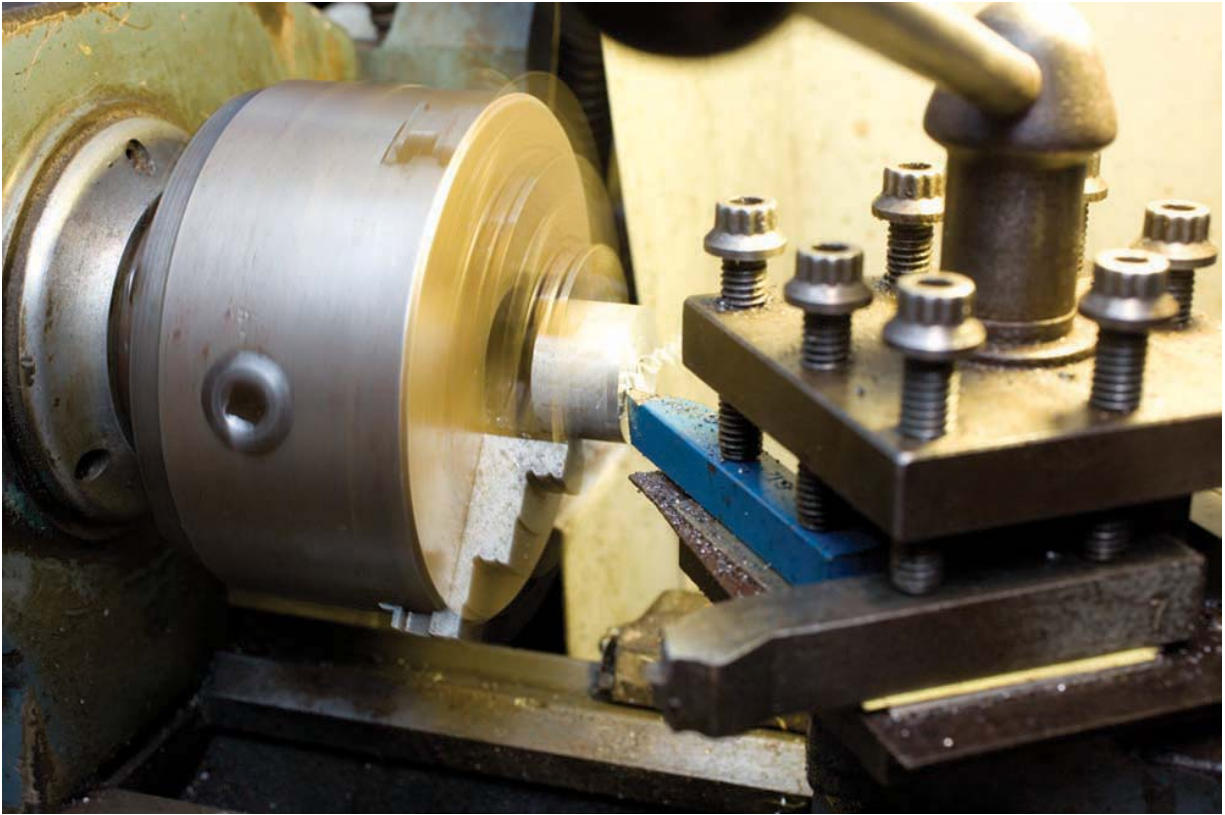


Photo 59 Favourable machinability, apart from long swarf trails, low strength aluminium alloy, facing-off in the lathe.

3 *Behaviour of metals*

STRENGTH OF MATERIALS: STRESS AND STRAIN

A handful of concepts in the field of engineering materials are fundamental to understanding the subject; these concepts form a continuous thread from the most simple to the most complex problem we tackle. Here we encounter the ubiquitous 'stress–strain curve'.

Wherever the behaviour of metals is considered, the first thing needed is a grasp (and sometimes just a general knowledge is sufficient) of this relationship between stress and strain. In the beginning was the simple tensile test.

Stress

The concept of stress is simply the load applied to a body divided by the area over which it is applied. Although a fundamental parameter, it is not directly visible in the material. It is normally introduced by examining the simple 'uniaxial' (load in one direction) tensile test. Applying load (F , as Newton or lbf) to the test bar of cross section area (A , as mm^2 or in^2) results in a stress (as N/mm^2 or lbf/in^2), as F/A .

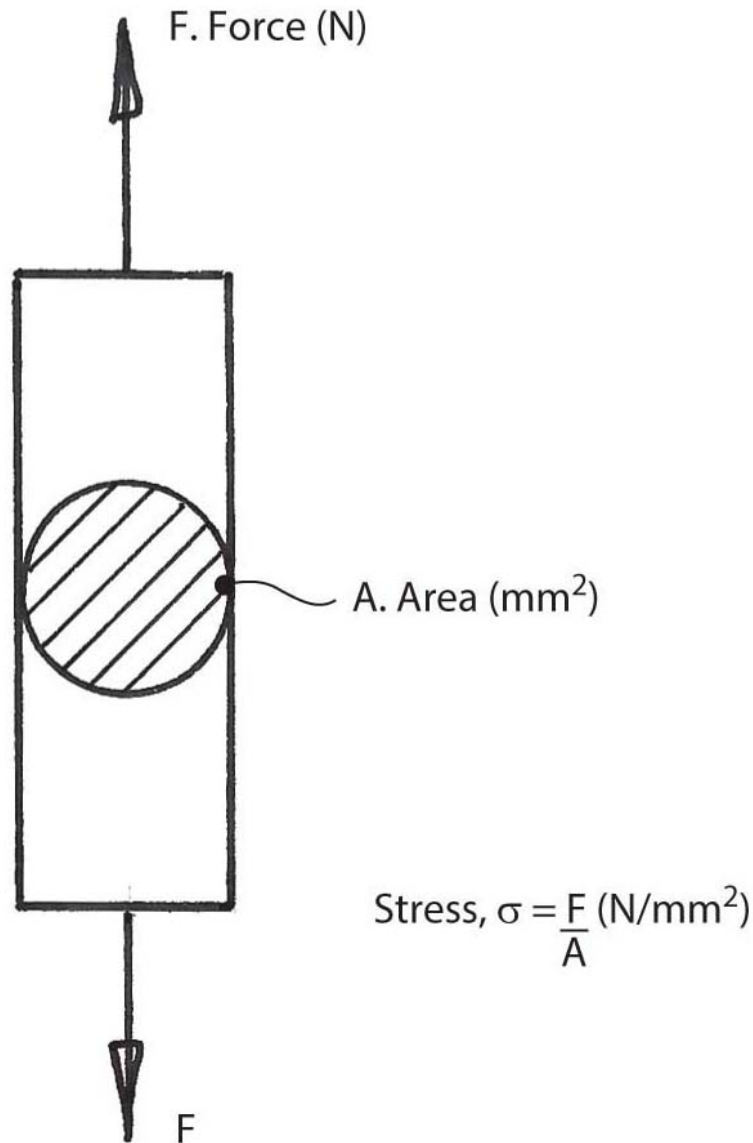
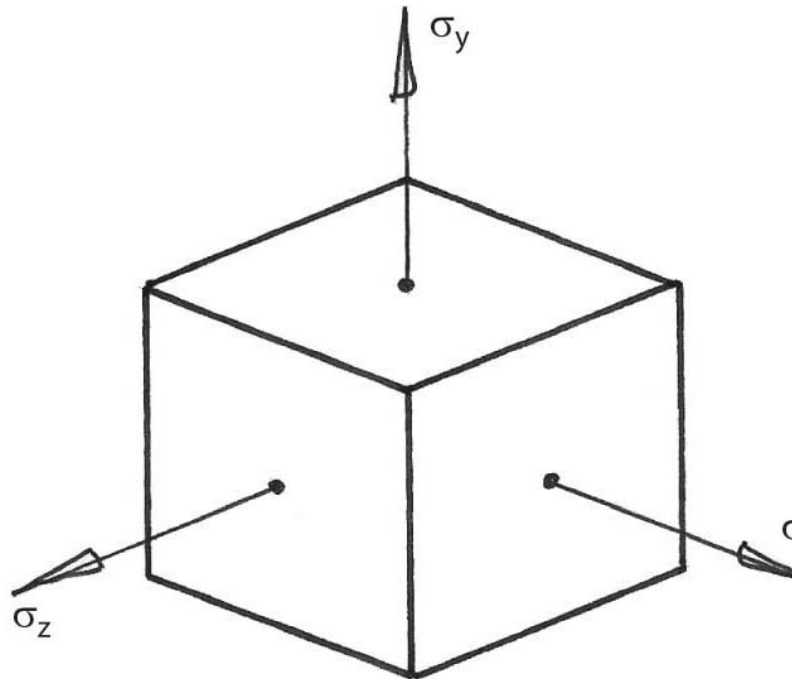


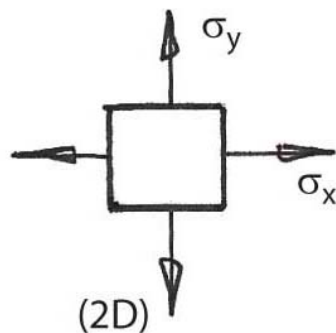
Figure 4 Stress.

Note: care must be observed in using units of force (F). Force is mass $\times$ acceleration due to gravity (g). In metric units, F (N, Newton) = mass (kg) $\times$ 9.81 (m/s²); and in Imperial units, F (lbf, pound force) = mass (lb) $\times$ 32.2 (ft/s²).

The single load applied in our tensile test provides an average stress over the whole section, but in practice we can use a simple stress element (see [Figure 5](#)). This can be used to examine stresses at a point, providing a very powerful tool in the understanding of the behaviour of real materials in action.



(3D)



(2D)

$$\varepsilon_x = \frac{\sigma_x}{E}$$

$$\varepsilon_y = -\nu \frac{\sigma_x}{E}$$

σ = Stress (N/mm²)

ε = Strain (mm/mm)

ν = Poisson's ratio (unit less)

Figure 5 Stress element.

Strain

The concept of strain can be considered in the context of the simple tensile test, defined as the *ratio* of extension divided by the original length, i.e. strain (ε) = $\Delta L/L$, where L is the original length and ΔL is the change in length due to the applied load. Fortunately, provided we work in consistent units of measurements (either mm or in), we benefit from strain being a unit-less parameter.

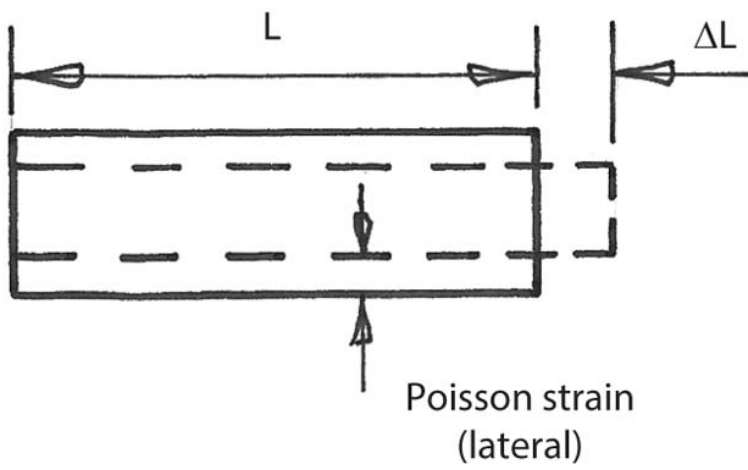
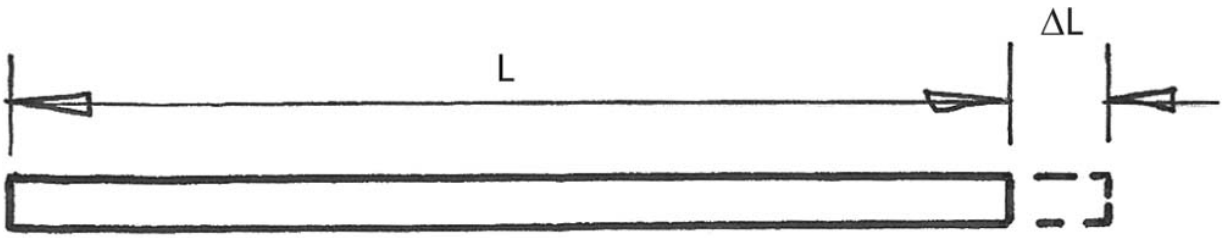
The stress–strain curve

For a bar of mild steel this is strictly speaking a load/original area vs. extension graph, universally used for engineering purposes. A *true* stress–strain curve is based on this, but with stress being load divided by the true cross-section area, which changes during testing. Notably, the test piece section will also reduce significantly after the peak of the load-extension curve. Extension is simply changed from length units (mm or in) to unit-less strain (mm/mm or in/in).

Hooke's Law

The first feature, and perhaps the most important of all, is the straight-line relationship from the origin to the yield point (σ_y). This derives from the work of Robert Hooke (1635–1703), whose theory of elasticity enables prediction of stress from the measurement of strain, as described later. It should be noted that strains in metals, up to yield, are normally very small and require extremely accurate measurement. Nevertheless, now within the amateur's capability, it presents the opportunity for experimental stress analysis, as described later.

In the elastic region, stress is directly proportional to strain. Loading and unloading is indefinitely repeatable, with the proviso that 'fatigue' can upset the life of a component. This major limitation is introduced shortly, and examined in more depth later (see [Chapter 4](#)), as fatigue is implicated in the vast majority of all engineering failures.



$$\text{Strain, } \epsilon = \frac{\Delta L}{L + \Delta L}$$

Figure 6 Strain.

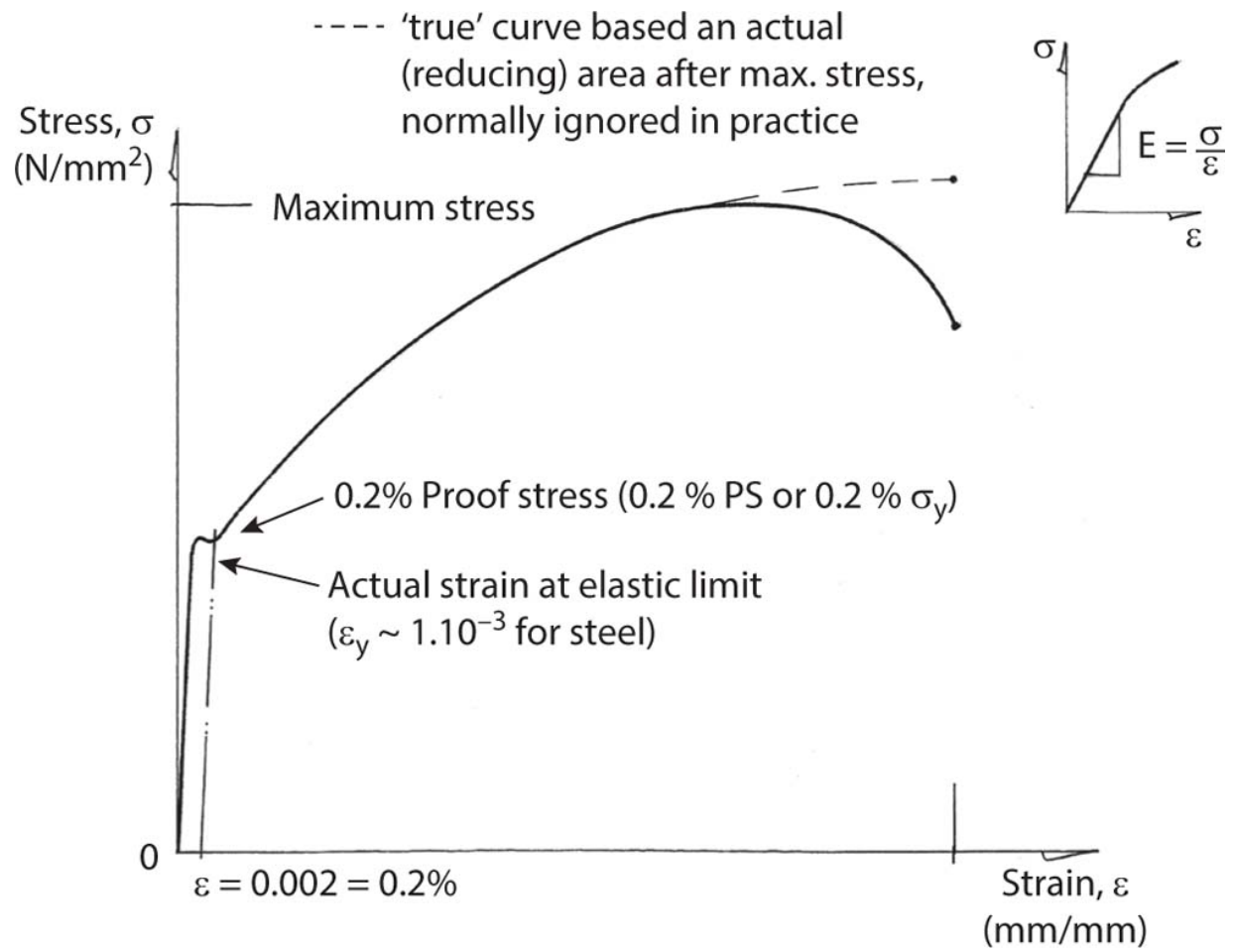


Figure 7 Stress-strain graph typical of mild steel.

Round Tensile ASTM 50mm Extensometer

Name	TestNo	Dia	Ext GL	Original GL	Proof Stress	Tester	Temp Deg C	CSA
ID		mm	mm	mm	%			mm ²
NO.1	78689a	12.52	50.00	50.00	0.20	tn	23	123.11

Name	Max Load	UTS	Proof Load	Proof Stress 0.2%	Elong	RA	Modulus (E)
ID	kN	MPa	kN	N/mm ²	%	%	GPa
NO.1	98.41	800	71.23	580	39.5	69.5	206

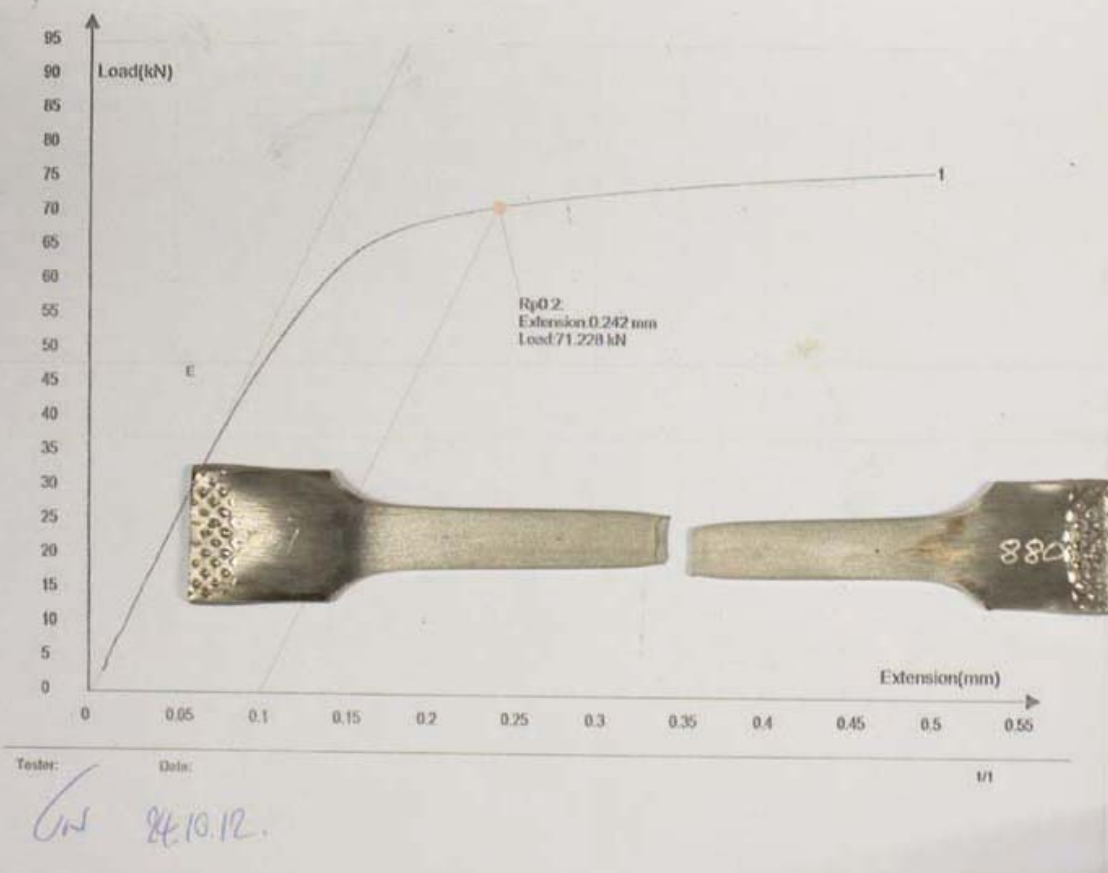


Photo 60 Tensile testpiece from commercial test house work.

Young's Modulus

This elastic region, with its linear relationship between stress and strain, produces another important characteristic, stiffness, as documented by another fine scientist, Thomas Young (1773–1829), who presented his eponymous 'Modulus of Elasticity' – (E) equals stress/strain. Thus, the slope of this part of the stress/strain graph provides E for mild steel as $\sim 270 \text{ MN/mm}^2$ or $\sim 30 \times 10^6 \text{ ksi}$, E being a constant for each alloy.

Table 20 Modulus of elasticity (E), rigidity (G) and Poisson's ratio (u).

Material	$E \times 10^3 \text{ ksi}$	$G \times 10^3 \text{ ksi}$	ν
W	58	22.8	0.27
steel (plain C and low alloy	29	11	0.33
stainless steel (18/8)	28	9.5	0.28
Ti	17	6.5	0.31
Cu	16	6	0.33
Al alloys	10.5	4	0.31
concrete	2		0.1
nylon	0.4		0.4
rubber	0.002		0.5



Photo 61 Tensile test machine, also possible to employ for uni-axial fatigue testing.



Photo 62 Tensile and creep testing machine, with elevated temperature facility.

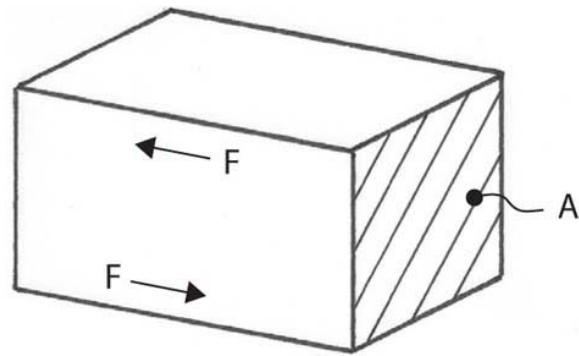
Poisson's Ratio

Returning to the uniaxial tensile test we find that as the test piece is extended in length, it shrinks in diameter, partially maintaining its volume. This ratio of longitudinal extension to lateral contraction is universally known as Poisson's Ratio (ν), which is around 0.3 for metals. As for E , ν is a constant for each material system, and enables further calculations in stress analysis, as shown in [Figure 22](#). The non-metals often behave in a different manner under tensile loading. In concrete, the value of ~ 0.1 shows that its lateral contraction is very low, whilst rubber (~ 0.5) undergoes a relatively large lateral change, indicating its virtually zero volume change.

Shear modulus

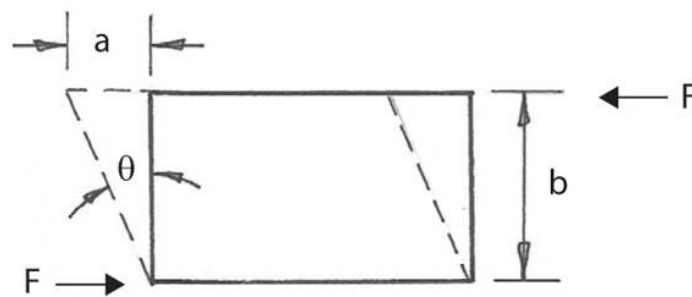
The material under analysis can, instead of being loaded uniaxially, be loaded in shear, causing a distortion of the cube, as shown in [Figure 8](#). This produces shear stress (τ), shear strain (γ), and the resultant shear modulus, the 'Modulus of Rigidity' (G), in a manner analogous to uniaxial loading with σ , ϵ and E . Thus $G = \tau/\gamma$, but this can also be found from our uniaxial parameters, providing the very useful relationship, $G = E/2(1 + \nu)$. As a handy rule-of-thumb for metals, $G \sim E/3$.

This concludes all the basic concepts for simple stress analysis, to provide a means of estimating the viability of a structure or component. But before progressing this, consideration needs to be given to what happens beyond the elastic limit of the stress–strain curve, the *plastic* region.



$$\text{Shear stress, } f = \frac{F}{A}$$

for shear, use 'Modulus of Rigidity', G ,
 so $\frac{\lambda}{\theta} = G$



$$\text{Shear strain, } \gamma = \tan. \frac{a}{b}$$

as small angles, $\tan \sim 1$

$$\text{so } \theta = \frac{a}{b}$$

Figure 8 Shear stress and strain.

Plasticity

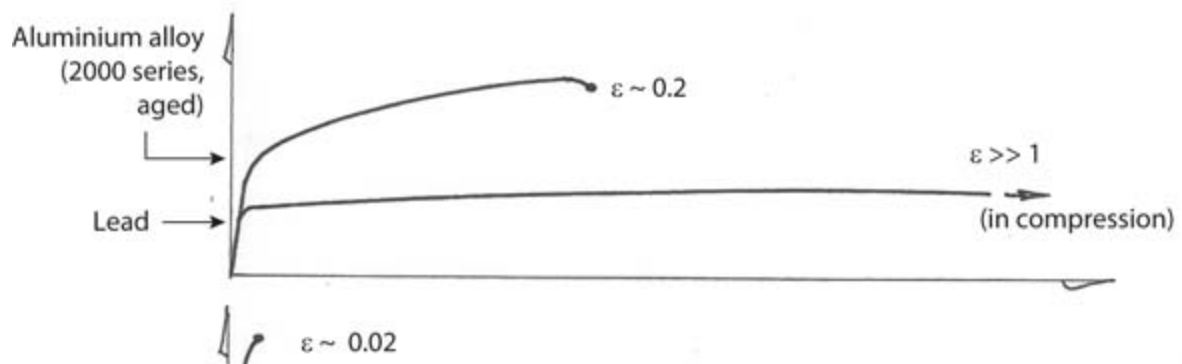
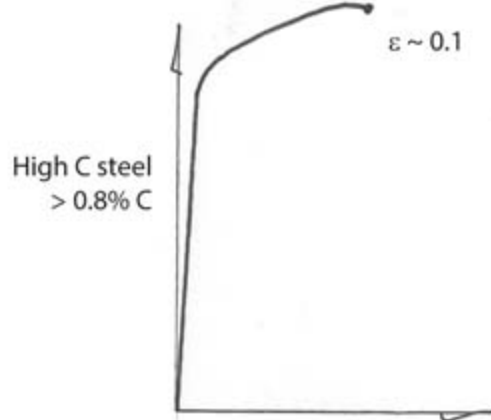
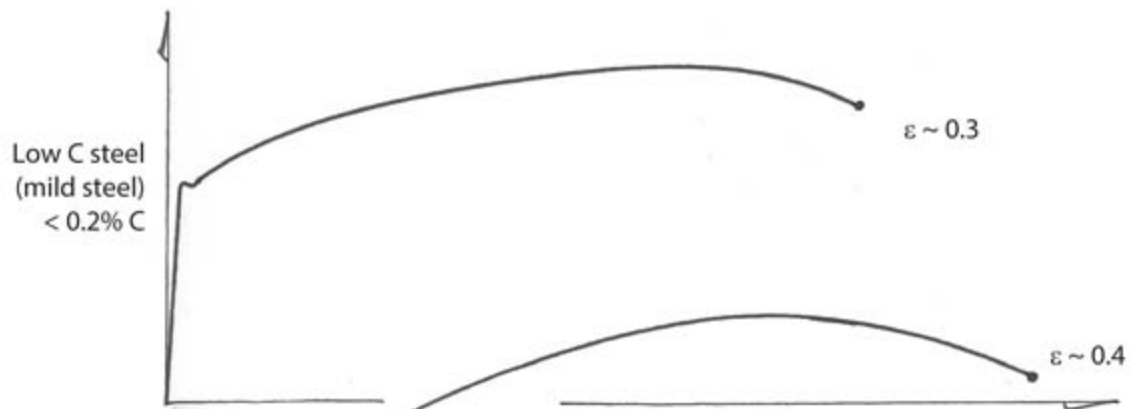
From the mild steel stress–strain curve, the region beyond the elastic limit is one of irrecoverability, where relaxing the load no longer returns the material to its previous dimension, but suffers a permanent set. This is something with which we are familiar in ductile materials, such as bending a stainless steel kitchen fork.

Before leaving the elastic region it should be noted that the elastic limit, or 'yield point', is rather tricky to measure reliably as a precise value in different metal systems. In mild steel there is normally a small dip after the yield point is reached, followed by the plastic region providing an upper and lower yield point, the former being used for yield strength. Thus, for consistency, it is normal to measure the yield strength as the stress at a prescribed strain, most commonly 0.2 per cent strain, quoted as '0.2 per cent PS', as shown in [Figure 7](#).

Due to the steepness of the elastic part of the graph, it can be seen that a small variation in establishing a yield point is to be anticipated. In practice, quoted figures for yield point (σ_y) need to be treated with caution, as this is a critical measure of the material's capability to support its design loading without permanent deformation. In the case of non-ductile materials, failure can occur with virtually zero plastic deformation in the stress–strain curve.

The stress–strain curves shown in [Figure 9](#) show the difference in plasticity typical of a variety of materials, from mild steel which demonstrates an elongation of ~30 per cent, to grey or white cast iron with virtually no plasticity, often failing at <1 per cent elongation.

The shape of the stress–strain curve in this region also provides a valuable insight into the work-hardening characteristics of the alloy. This arises from the slope of the curve from yield to ultimate stress; the steeper and higher the curve, the greater the rate and extent of work-hardening. However, this increased strength is produced at the cost of reduced ductility. The load versus extension graph shows a peak, followed by reducing load to failure, although the material continues to work-harden as the section reduces by wasting, so that the true stress–strain graph would retain a continuously rising curve to final failure. However, the onset of necking is a point of significant instability and in practice it would rapidly presage failure. In ductile materials, this reduction of area ('R of A') is also used as a measure of ductility, as is elongation, where R of A can be 50 per cent for mild steel.



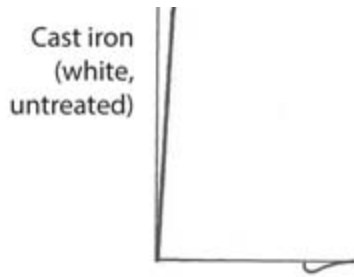


Figure 9 Stress-strain curves (schematic and generalized).

Austenitic stainless steels (e.g. grade 304) and 70/30 brass have notably high rates of work-hardening, combined with high ductility. This is an important property, for these alloys are not susceptible to hardening by the usual means of heat treatment. Plastic deformation is often deliberately employed in high integrity bolting where careful overloading, beyond yield, produces improved strength. Naturally, this has to be closely controlled in order to avoid failure.

In other alloys the work-hardening rate is low, which is beneficial in forming operations, but provides less safety margin if yield is exceeded unintentionally. In this case a relatively small load could then progress the piece to ductile failure. This illustrates another virtue of ductile steels, which have a high ratio of ultimate stress to yield stress; typically, in mild steel, ultimate stress $\sim$ twice yield stress.

A small number of materials, such as lead, have a work hardening index of around zero, so that increasing load, after yield, produces extension with virtually no increased load. This can present practical problems in engineering applications, resulting in severe 'creep' or increasing extension, or strain at a constant load. Creep can be a common phenomenon at high temperatures, and may be design-limiting in critical parts like the jet turbine blade, but at room temperature it is limited to materials like lead, or in 'super-forming' materials at extremely low strain rates. As noted earlier, in material processing, this enables lead to be used when modelling deformation processes, such as rolling, where its room temperature characteristics *in compression* replicate the stress-strain behaviour of practical engineering materials at high temperature.

Whilst stress analysis has traditionally been limited to the elastic region, where stress is proportional to strain and safe limits to design can generally be set to ensure freedom from failure, it is also possible to extend this analysis into the plastic region. However, the game then becomes vastly more complicated and for our purposes this can be treated as of purely academic interest. Incorporating significant plasticity is vitally important when designing for plastic collapse, allowing 'shakedown' overloading safeguards, and has seen tremendous benefits in vehicle crash resistance, ultimately improving road and racing vehicles.

Early work on plastic collapse by Professor John Baker of Cambridge University led to the design and highly successful application of the Morrison Shelter, to protect civilians from house collapse during bombing of UK cities in the Second World War. An analysis later showed that these devices achieved over 95 per cent success in saving lives through the controlled plastic collapse of the simple 'table-like' structure, absorbing the energy of the falling house. Tests confirmed the designed collapse of ~300mm on impact. A self-assembly pack provided four 1m mild steel legs and a base, supporting a 3mm-thick mild steel plate, about 2m², like a table-top, under which a family could sleep with relative safety during bombing.



Photo 63 Pedestrian protection railings, showing impact absorbed by vehicle strike, safe plastic deformation of mild steel, even retaining galvanizing corrosion protection, long after damage!

The need for a better grasp of elastic and plastic stress analysis has been the driving force for the understanding of material performance, at temperature and under fatigue and creep conditions, where design is critical and safety factors approach unity, in weight-critical vehicles like aircraft. Safety factors (SF) include the ratio of operating stress to yield stress, of importance in theoretical

analysis of all deformation processes. Commercial Finite Element Analysis (FEA) now enables computer modelling, incorporating plastic deformation, for structures that were previously beyond practical calculation. For our purposes, however, the elastic stress analysis should be quite sufficient.

Failures, whilst obviously disastrous, can provide invaluable insights into the behaviour and limitations of engineering materials and their application in the real world, compared to the theoretical one. (See 'BOAC Flight 781' below.)

The study of the Comet disasters illuminates aspects also relevant to smaller-scale operations: design by stress analysis, verified by strain gauge measurement; the use of elastic behaviour subject to variation by plastic deformation (shakedown); stress concentrations due to section changes and holes; residual stress; and all of the factors that may not critically affect components subjected to single loading cycles, but in fatigue can lead to failure at only half the yield strength.

BOAC FLIGHT 781: FAILURE ANALYSIS

The British Comet disasters of the early 1950s are amongst the best known engineering failures of modern times and many are familiar with the cause – attributed to fatigue cracking from the fuselage windows' corners. Whilst this is generally correct, the full story provides valuable insight in relation to engineering materials and their application.

This part of the tragic tale begins on 10 January 1954 with BOAC Comet 1, GALYP, taking off at 09.34 GMT from Rome on its final leg from Singapore to London, with twenty-nine passengers and six crew aboard. At around 09.51 hrs GMT, still climbing at around 8000m, a passing BOAC Argonaut G-ALHJ received a radio call from the 31-year-old Captain Alan Gibson in the Comet: 'George How Jig from George Yoke Peter, did you get my. . .', violently interrupted as the Comet suffered a sudden explosion and subsequent crash into the Mediterranean Sea, just off the Island of Elba, with the loss of all aboard.

This unexplained catastrophe took the world by surprise as the Comet was the first commercial jet manufactured by the renowned firm of de Havilland in England, benefiting from the Whittle jet engine development that finally reached fruition in the English Midlands during the latter part of the Second World War.

The Comets began service on 1 May 1952 after extensive development but, with hindsight, insufficient testing to evaluate the dynamic loading of the main structure. The Comet 1 was powered by four de Havilland Ghost engines of 22,000 N thrust, providing a payload of fifty passengers. By 1990 the Rolls Royce RB211-254G engines were producing ten times the power (250,000 N) with a payload of 450

passengers. Although the Comet 1's engines appear low-powered compared to present versions, they were a dramatic advance over the turbo-prop propulsion that preceded it, raising the flight ceiling from ~5,000m to ~12,000m at a stroke. This had an equally dramatic effect on fuselage differential pressure, as cabin pressure is maintained at ~0.8 atm. At full cruising height the external pressure is ~0.2 atm, producing a differential pressure of ~0.6 atm. In effect, the fuselage is a large pressure vessel, basically a long cylinder.

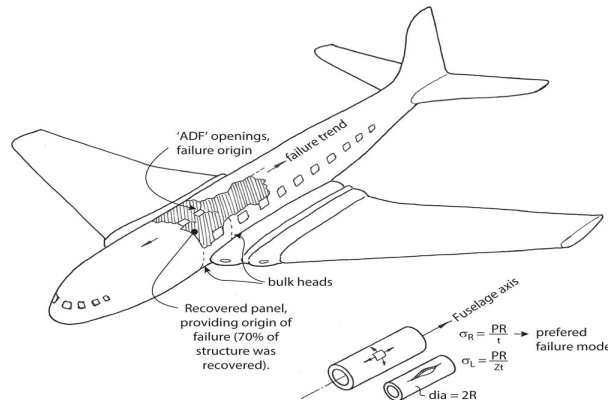


Figure 10 Comet 1, code G-ALYP.

After the crash of G-ALYP, there was great difficulty in understanding the cause of failure, as this preceded the **introduction** of 'Black Box' flight recorders, there was no well-established protocol for failure investigations, and the wreckage was submerged in the Mediterranean. Therefore, all Comet 1 flights were immediately cancelled and the entire fleet of British- and French-operated Comets was exhaustively examined, resulting in some sixty structural parts, thought suspect, being reinforced. Under severe commercial pressure to resume service, the British Aviation authority raised no objections and on 23 March 1954 BOAC resumed their service with the Comets.

It was recognized that G-ALHP had completed only 3,600 flight hours, equating to 1,290 flight cycles, where the fuselage undergoes a pressurization cycle for every full ascent and descent. The design life predicted by the de Havilland engineers was around ten times that undergone by this plane and clearly a factor in re-starting the service. BOAC's Chairman stated on television, 'We obviously wouldn't be flying the Comet with passengers if we weren't satisfied the conditions were suitable.'

But then, on 8 April 1954, a BOAC Comet G-ALYU on a charter flight for South African Airways took off from Rome, bound for Egypt. Some thirty-three minutes after take off, at around 10,000m, all contact was lost and the wreckage found the next day indicated that the plane had gone down in the 1,000m-deep waters of the Bay of Naples. All fourteen passengers and seven crew were lost, and a dazed BOAC grounded the whole fleet of Comets until a proper understanding of these disasters was found; both flights had been undertaken in good weather, and there was sufficient evidence from debris recovered to rule out sabotage. A structural weakness was now highly suspect, despite the predicted design life.

Efforts were re-doubled to retrieve sufficient fuselage and structure from the first failure, as G-ALYP had gone down in 180m-deep waters – deep for contemporary sub-sea investigation but far less than the second crash. Two British Naval vessels

made use of an innovative underwater inspection chamber with first-time use of CCTV and managed to recover around 70 per cent of the superstructure, 50 per cent of equipment, and 80 per cent of the engines of G-ALYP. The wreckage was re-constructed by an expert team at the Royal Aircraft Establishment (RAE), Farnborough, England, who found that the fatal cracks had begun near the automatic direction finding (ADF) windows, located on the top of the fuselage, between two bulkheads, above the leading edge of the wings. At the time of writing this crucial piece was exhibited in the Science Museum, London.

The cracks had progressed along the length of the fuselage, typical in pressure failures of internally pressurized cylinders, as radial stress is twice that of longitudinal stress in a thin-walled vessel. When these cracks reached the bulkheads they turned through 90°, causing a flap to rip up like the opening of a tin can. Driven by the explosive force of the cabin pressure, it led to rapid disintegration of the whole structure.

The problem was in reconciling the clear mismatch between the theoretical strength calculations and the early failure of both aircraft, at 1,290 and 900 flights against the predicted life of 10,000 flights. This is not an unusual state of affairs when investigating failures of industrial products such as rotating-shaft pumps, where experimental data is sparse and understanding the failure is limited by insufficiently developed theory, but in critical structural items like aircraft the theory is expected to be far more reliable.

In order to recover a disastrous situation, BOAC provided the complete airframe of another Comet, G-ALYU, for the Royal Aircraft Establishment at Farnborough, and pressure cycling began on a full-scale test to replicate the flight history of these aircraft, in the days long before computer modelling was developed for stress analysis of such complex structures. G-ALYU had already made 1,230 flights when it began simulation testing, which had to be done by immersing the whole structure in a water tank whilst also pressurizing the cabin internally with water, as air pressurization would release, on failure, energy equivalent to a 225kg bomb! (Failure during a hydraulic (water) test merely results in sudden leakage, easily contained, and the energy is safely dissipated.)

These tests cycled the cabin from zero to normal operating pressure, establishing the effects of fatigue to test the prediction of 10,000 flights before failure. In the event, severe fatigue cracking occurred after only 1,830 cycles, making the whole life of G-ALYU 1,230 plus 1,830 flights. At 3,060 flights, this was far lower than predicted.

The fuselage skin comprised 'DTD-546' sheet, an aluminium alloy no longer available, but of the 2000 series. Yield and tensile strengths were taken as ~350 and 450 MN/m², respectively. Hand calculations by de Havilland indicated stresses at the window corners of ~195 MN/m². The International Civil Aviation Organization (ICAO) set the standard of the day for fuselage design: firstly, ability to withstand 33 per cent above normal cabin pressure without permanent deformation; secondly, testing to 100 per cent above normal pressure without exceeding the ultimate tensile strength. This latter case would result in a calculation value of 390 MN/m², safely within the material's strength and the ICAO rules.

During the earlier design calculations (prior to the crashes), de Havilland decided to 'play safe' by increasing the design pressure from 100 to 150 per cent of normal. They prepared a section of fuselage for testing and evaluation, and subjected it to two 100 per cent pressure tests without any problems. But it was not until July 1953 (after the Comets had been in service over a year!) that de Havilland decided to explore the possible effects of fatigue. Although the civil authorities had not yet

anticipated this as a problem, there was growing awareness in the industry of the danger of fatigue, although it was not yet properly understood in this context. Presciently, Nevil Shute (Norway) fictionalized structural fatigue damage, before the Comet failures, in his novel *No Highway*, although he reflected the metallurgical misconceptions prevalent at the time (and persisting in some areas!) about the mechanism of fatigue, as discussed later (see [Chapter 4](#)).

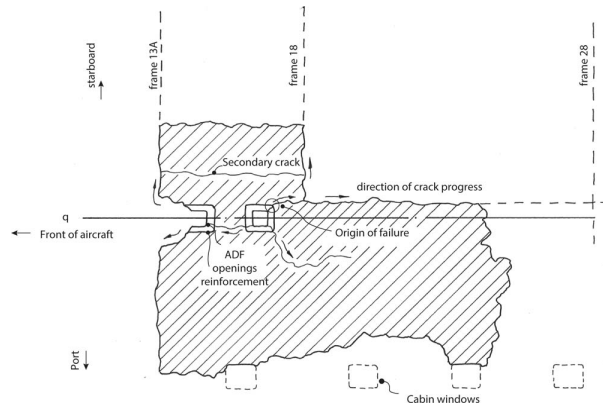


Figure 11 Recovered panel from Comet G-ALYP.

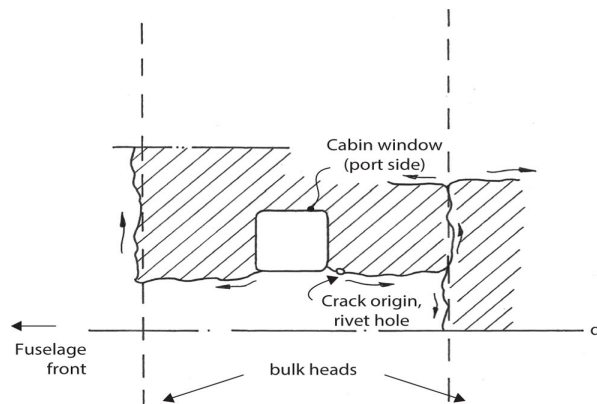


Figure 12 Origin of failure, G-ALYU, simulation test at RAE, Farnborough.

This de Havilland test-section was cycled between zero and normal pressure and achieved some 18,000 cycles before failing by fatigue cracking from a defect near a window corner. This appeared to show that the design life of 10,000 flights was amply validated, and supported the maker's confidence in their design, before the two disasters of 1954. Returning to the full-scale fuselage test of G-ALYU, [Figure 12](#) shows how the cracks had progressed after the total life of the 3,060 cycles.

These cracks were from an escape hatch, just forward of the wing, and progressed in a similar manner to that on the first Comet failure, G-ALYP. Extending to the adjacent bulkheads, they turned through 90° and began the tearing mechanism, relieved in the test by escape of the pressuring water, but which would have produced an immediate and catastrophic explosion in flight.

Before complete rupture, the fuselage was then patched up to enable strain gauges to be fixed to the critical area, and the cabin re-pressurized. The maximum stress appeared near the window corners, at a value of 197 MN/m^2 . Whilst the crack

origin was at a rivet hole, beyond the apparent maximum stress area, this would have been subject to significant stress concentration. Additionally, there were small-scale manufacturing defects associated with discontinuities, like the rivet holes.

All of the required data was now available, although by now considerable time had elapsed since the Comet fleet had been grounded. The official enquiry also assessed the earlier part-section testing by de Havilland, in 1953, which had provided the result of an 18,000-cycle life. This model had been subjected to the additional 'safeguards' of over-pressurization tests of 100 per cent, whereas the production aircraft, like G-ALYP, were subjected only to a single 33 per cent over-pressure test. Then, during the part-section testing, the fatigue test was interrupted at intervals to conduct further over-pressure tests, whereas the full-scale fuselage of G-ALYU was simply cycled continuously until failure.

This was the first critical finding, that the over-pressure testing had over-stressed the aluminium panels around the rivet holes due to their stress concentration, causing local plastic deformation. This produced an increased local strength and, on unloading, a compressive residual stress, thereby greatly improving the resistance of this critical area to fatigue crack initiation and propagation. Therefore, somewhat counter-intuitively, the over-loaded test unit produced a structure with significantly enhanced fatigue life, compared to the production airframe which had been subject to just a single lower-pressure test prior to entering service.

The second major consideration concerned the weight given to the results from the single, part-section, fatigue test of 1953. Even at that time, it was well known that fatigue testing is subject to significant variability and therefore a single test result should not have been taken as a reliable predictor of this critical parameter. A statistical basis for experimentation should have required a series of tests to have been conducted, for unbiased confidence levels to be established.

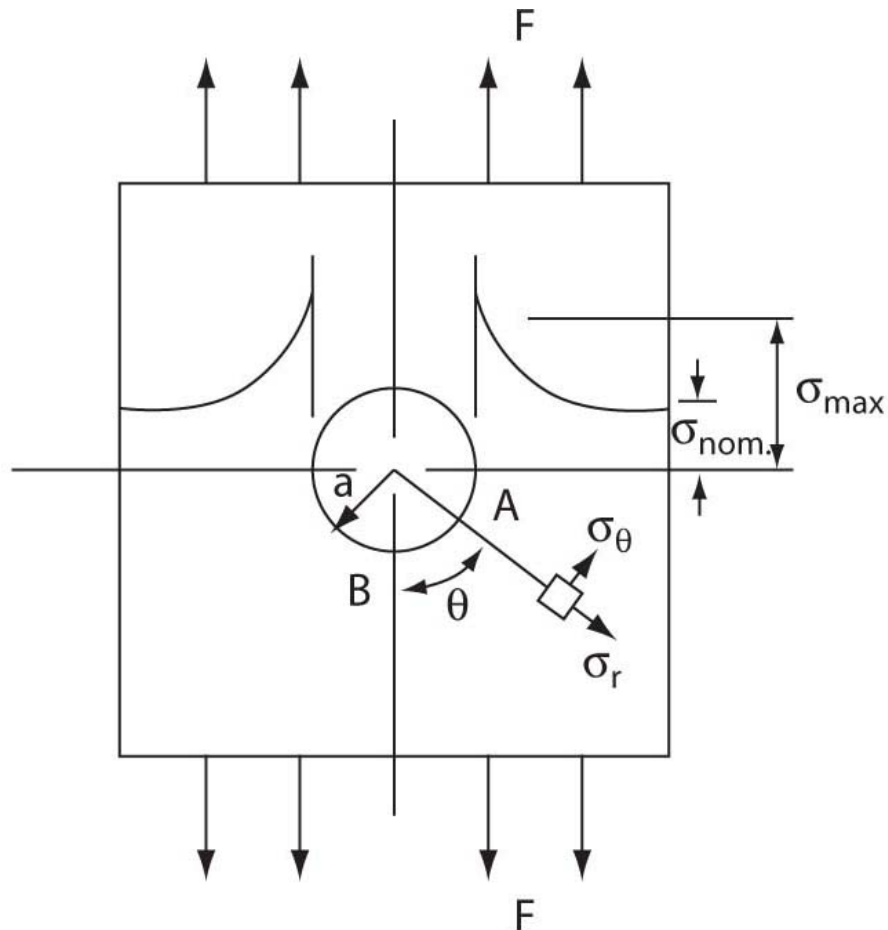
Ironically, the rivet holes near the windows and port holes that provided the origin of the fatal cracks were introduced only for expediency of affixing the doublers and stiffeners around these apertures. Thus, riveting (instead of using adhesives as done elsewhere on the structure – a method pioneered by de Havilland) produced a stress-concentration around already highly stressed material. And this was not appreciated at the time by the designers.

A consequence of the elapsed time during the investigation was that the great innovation-lead of Britain and de Havilland was lost, as the USA and Boeing developed and successfully introduced the Boeing 707. By the time the greatly improved Comet 4 returned, the commercial advantage could not be regained and the USA manufacturers led the world in commercial aircraft for the remainder of the twentieth century, until the advent of the European Airbus.

Stress concentrations

Wherever there is a geometrical discontinuity, such as change in section, profile, hole or gross defect, there will be a 'stress concentration' (K_t). This is usually described as follows: K_t maximum stress/nominal stress.

The classic case is the 'hole-in-plate' analysis, which is actually what we met in the Comet failure review. As usual, the elegant maths has been done long ago and we don't have to follow this from first principles to benefit from the result (although those who are interested should see the standard texts listed in the [Bibliography](#)). From [Figure 13](#) we can see that the hole produces a stress concentration, $K_t = 3$. The surface of a hole in a flat plate produces local tensile stress that can rise to three times the nominal stress, where nominal stress is the load divided by the remaining cross sectional area supporting the load. Interestingly, the same value of stress is applied at point b, but here it is *compressive* and so normally not significant.



- (B) Stress at $\theta = 0^\circ$
i.e. $\sigma_{\text{nominal}} \left(\frac{F}{\text{Area}} \right)$
- (A) Stress at $\theta = 90^\circ$
i.e. $3\sigma_{\text{nominal}} = \sigma_{\text{max}}$

Figure 13 Stress concentration, circular hole.

For the case of an elliptical hole, we have $\sigma_{\text{max}} = \sigma_{\text{nom}} (1 + 2 a/b)$, as shown in [Figure 14](#). For the circular hole, this reduces to $a = b$, providing the same result as before, i.e. $\sigma_{\text{max}} = 3 \sigma_{\text{nom}}$. But it is clear that as the ellipse lengths change with $a \gg b$, then the stress concentration increases significantly. In the ultimate case of a sharp notch, a fatigue crack, the crack tip stress concentration rises dramatically, providing the dangerous condition that can drive forward a crack at deceptively low stress.

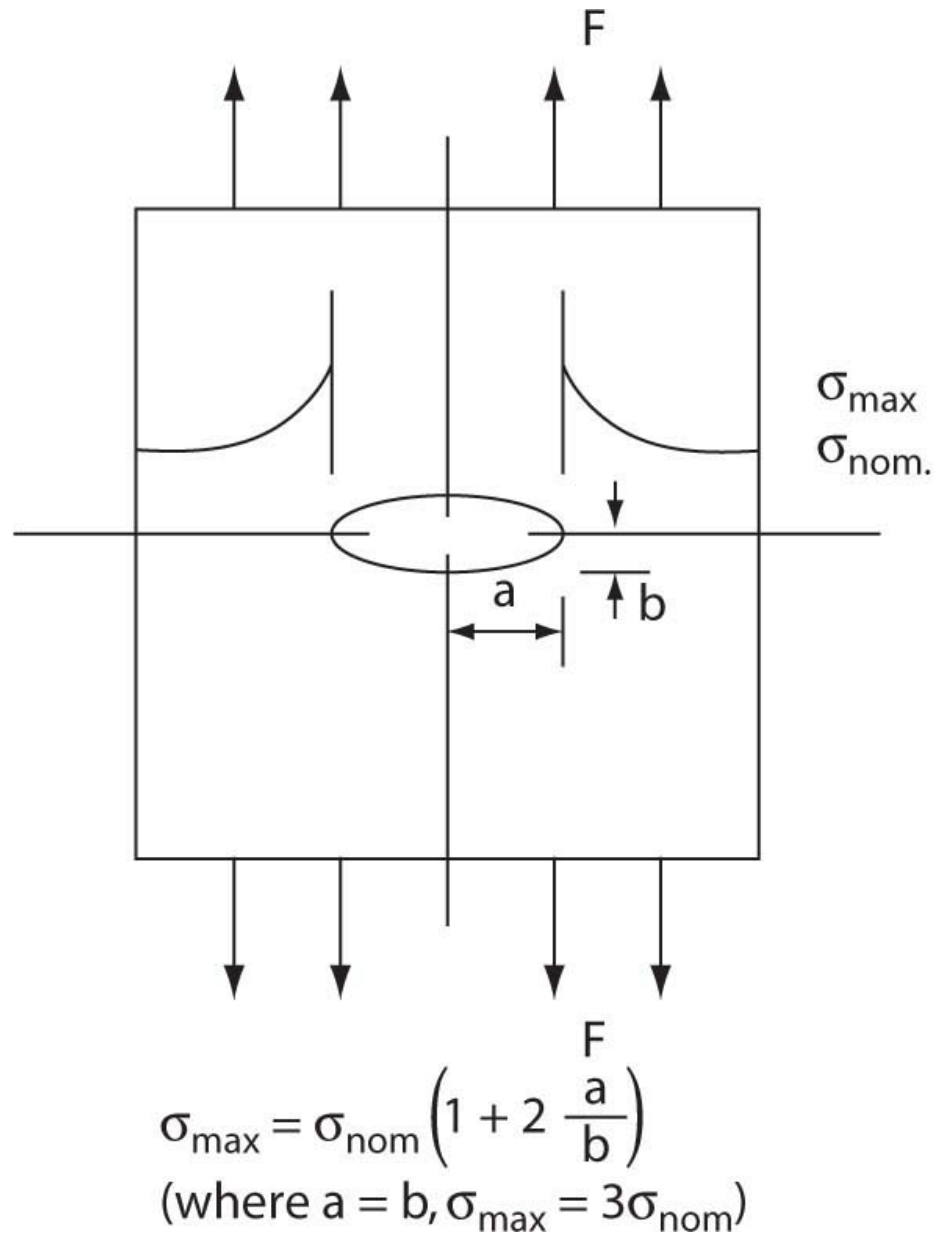


Figure 14 Stress concentration, elliptical hole.

Fortunately, ductile materials normally form the basis of structural members. Macroscopic plasticity in the region of maximum stress, or microscopic plasticity around the sharp crack tip, blunts the effect, enabling relaxation of local strain and improved stress distribution.

With Finite Element Analysis (FEA) and modern computational power, it is relatively easy to see the effects of these local stress distributions and, where repeated cycling does *not* occur, localized

high-stress regions can normally be tolerated, producing a 'stress shakedown' condition which can actually enhance the structural behaviour. Indeed it sometimes seems, in industrial applications, that FEA models can tolerate local overstressing almost without limit. In the Comet stress analysis this was a crucial element, the key point being that whilst this may be acceptable for occasional loading, the same is not the case under fatigue conditions unless the beneficial effect of residual compressive stress is applied. This was the case in the earlier Comet tests, which were subjected – unwittingly – to stress shakedown and a beneficial residual stress (see [next section](#)).

Where brittle materials are employed, then the effects of stress concentrations, even in single load, need to be handled with great care. Normally, this applies to tensile loading, so that grey cast iron can fail suddenly if loaded in tension unless stresses are kept sufficiently low to accommodate the inevitable stress concentrations common in complex cast shapes. Whilst this is understood intuitively by the experienced engineer, and non-ductile materials are used predominantly in compression or lightly stressed in tension, there are conditions such as 'tri-axial tension' and 'rapid loading' that can lead normally ductile materials to behave in a seriously non-ductile manner. This is examined later, where we see that the analysis assumed so far, in thin material, can break down and be rendered unsafe with increased thickness, use below critical temperatures, or at high strain rates.

Thus, stress concentrations and high strain rates are deliberately used in impact and fracture toughness testing, to evaluate the likely ductility of the material in particular applications, as discussed further in [Chapter 4](#).

The stress concentrations for the normal changes in section, such as grooves and shoulders on shafts, bars with steps, gear teeth and keyways, can be found in reference works (see [Bibliography](#)). Screw threads on bolts and studs are particularly difficult to analyse accurately (see also [Chapter 4](#)) and a high safety factor is appropriate if in doubt. These are particularly important where fatigue conditions apply. A high local stress, safe for single loading,

can be sufficient to initiate a crack in components such as rotating shafts. Similarly, minute surface effects from machining or surface treatments can have a metallurgical or mechanical stress concentration and a profound effect on fatigue behaviour.

In general, stress concentrations appear almost everywhere in components, including welded joints, but are often less damaging than might be expected in the ductile materials encountered in structures. However, care must be employed when they are encountered in less ductile materials or where ductile materials are compromised by thicker sections, lower temperatures, or environmental effects like corrosion.

Residual stress

Residual stress is, surely, the elephant in the Engineering Materials room. Something of an enigma, virtually omnipresent yet invisible, it is the hidden force in countless engineering failures and has effects across the macroscopic and microscopic scale.

Difficult to measure and even more difficult to model in an entirely realistic way, even with modern computational power it can still defy accurate evaluation. Almost by definition, it is hidden, as it applies non-uniformly through the section so that direct strain measurement on the surface cannot be employed, in the usual manner for stress analysis from external loading.

Macroscopic effects

On the macroscopic scale, residual stress appears in the early stages of material processing. [Figure 15](#) shows a rolling process where the deformation is not uniform across the plate thickness. The compressive stress due to this plastic deformation is applied only to the outer fibres of the material and, as the material must be in a state of equilibrium, this is balanced by an equal and opposite tensile stress across the inner part of the section. Note that residual stresses are purely elastic and always limited to the yield stress.

Beyond yield, they are limited by the plastic flow that produces permanent deformation.

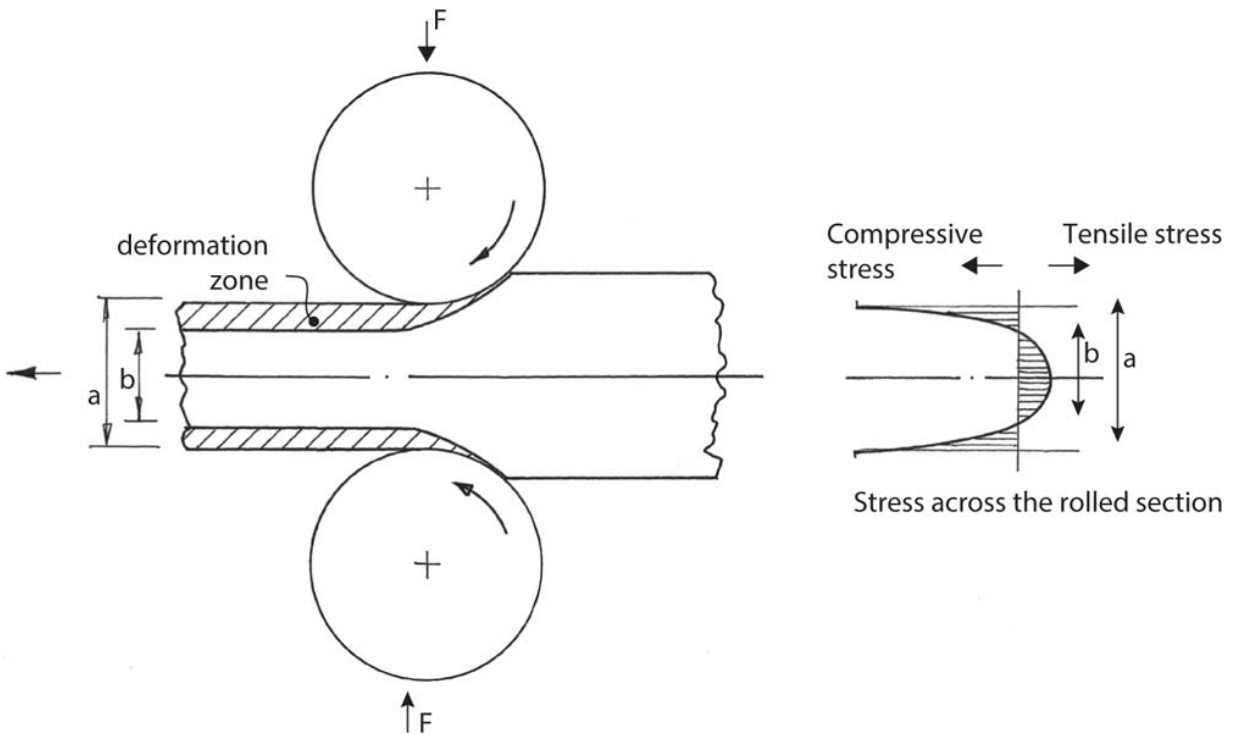


Figure 15 Residual stress rolling.

All forming processes, unless they are uniformly applied across the material section, will impose a degree of residual stress. As forming normally involves compressive loading, as for the rolling example, a similar state of surface compressive versus internal tensile stress can exist. Fortunately, this is generally a beneficial state of affairs in ductile materials, tending to protect the surface material from some degree of the tensile stress, which is normally a precursor to the onset of failures.

Another source of inherent residual stress arises from casting or heat treatment of an ingot. The variation in cooling rates due to the section size, where the core remains at a higher temperature than the outside, leads to non-uniform stress. Thus, the yield strength of the core may allow the thermal shrinkage of the outer material to compress the core, finally cooling to produce the residual stress shown in [Figure 16](#).

A similar situation applies to rapidly cooled (quenched) material during heat treatment, as shown in [Figure 17](#). This can be so severe as to contribute to early failure of a part, even before being put into service. Whilst the residual stress patterns generally follow those of the above diagrams, real life is often more complex. Tensile residual stresses encountered on the surface, often combined with stress concentrations, can be highly damaging in high-strength, heat-treated materials like low alloy steels.

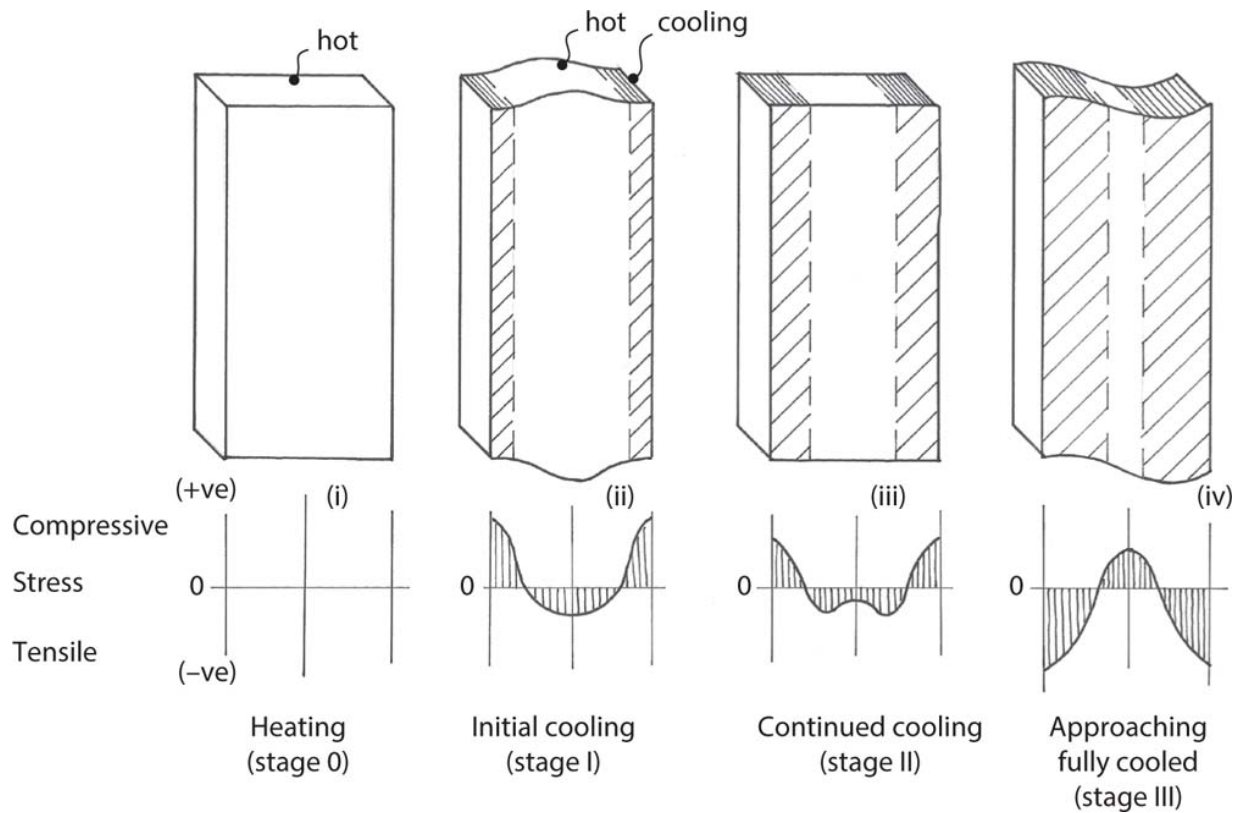


Figure 16 Ingot cooling stresses (based on Baldwin, W.M., *proc ASTM*, 40, 1949).

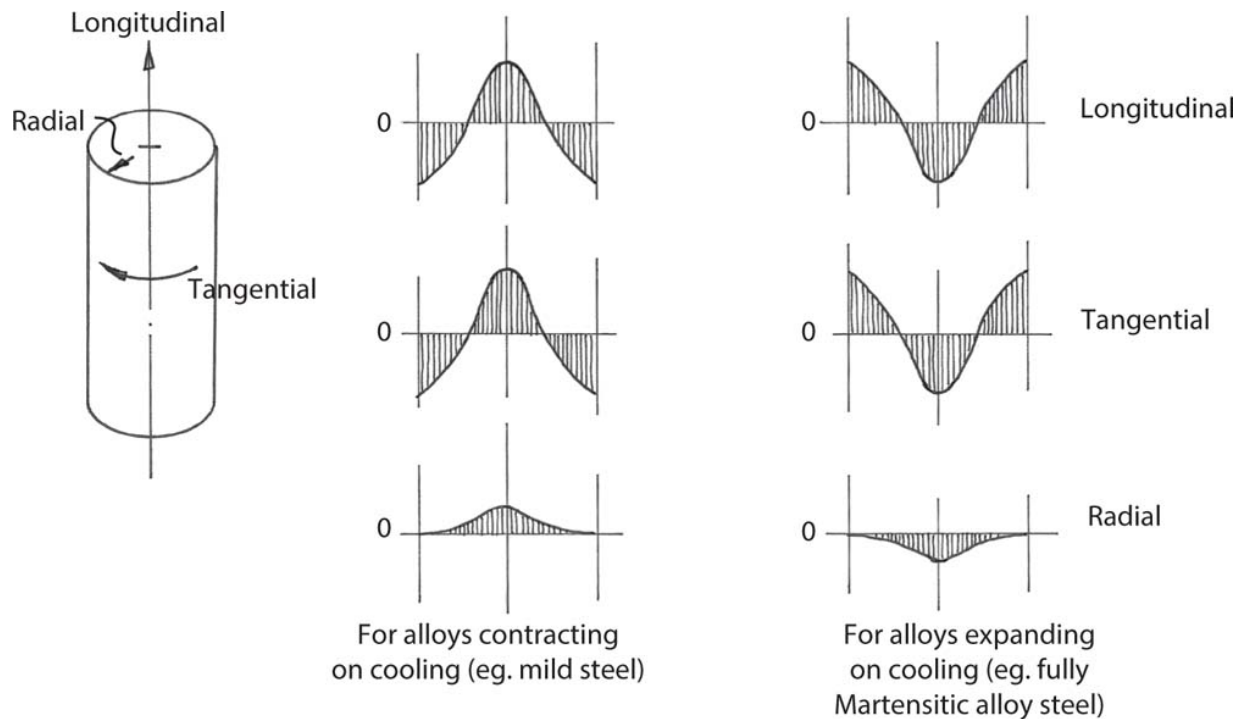


Figure 17 Residual stress in quenched pieces.

Machining, as a deformation process, is often a further contributor to residual stress. These effects, from casting, forging, heat treatment, machining and finishing processes (such as electroplating), are all likely to contribute to residual stress, to a greater or lesser extent. However, as evaluation of residual stress for each process is impractical, the options are either measurement in the finished product; stress relieving by thermal or mechanical treatment; destructive testing of a sample; or simply living with the possibility of unwanted and unknown residual stress. In practice the latter is often the case by default, and another example of how ductile materials, like mild steel, may save us from such hidden and unknown dangers by local plastic relief of the tensile 'hot spots' of internal (residual) stress.

Microscopic effects

Microscopic residual stress arises from cooling through phase transformations in the solid state, to room temperature. Particularly

important is the martensitic transformation in steels, responsible for the strengthening process, more of which in [Chapter 5](#). An oft-neglected result of this transformation, in steels, is the ~5 per cent volume increase where the full change takes place. In practice, this often does not occur uniformly throughout the section. Typically, the outer layers transform to near 100 per cent martensite (β'), whereas the core may be nearer to 0 per cent martensite, depending on the alloy characteristics. For example, an EN8 bar of 20 mm diameter, when air cooled, does not cool the core fast enough for complete transformation into martensite. Low alloy steels such as EN24, however, would transform more uniformly across the same section, and so would produce less residual stress. This effect can be used to advantage, as shown in [Figure 18](#), with the added benefit of the faster cooling rate possible with the smaller piece. [Figure 19](#) shows the tensile residual stresses that may occur in the part-transformed larger piece, through the section.

However, the most significant microscopic effect is the internal strain caused by the martensitic transformation. For the high strength and tool steels with carbon ~1 per cent and yield stress near fracture stress, a fully martensitic structure can provide sufficient internal strain to initiate cracks upon quenching (called, unsurprisingly, quench cracks). For these alloys it is necessary to modify the quenching process by reducing the severity, such as a slower cooling rate using oil rather than water; or interrupting the quench, to produce different transformation products, such as 'lower bainite'. This is the basis of 'austempering' (see [Chapter 5](#)).

For most applications, a tempering process is applied after quenching, and this re-heating relieves some of the internal transformation stresses, reflected in reduced strength and increased ductility, as discussed in [Chapter 5](#).

Note that this transformation behaviour is particular to the steels, and similar martensites do not necessarily occur in other alloys' systems. For example, some copper-base alloys transform to a *reversible* martensite, without producing the hardening and residual stress characteristic of heat-treated steels.

In age-hardening alloy systems, such as the aluminium and titaniums, the residual stress created by the ageing process is generally less of a problem, because heat treatment times are longer and through-thickness transformation the norm. Deformation, such as press-forming a sheet, tends to have a more predictable outcome, as it is more common to undertake the deformation before the component has been fully heat-treated (aged) to full strength.

Residual stress effects

As noted earlier, the primary importance of residual stress is whether it adds to, or subtracts from, the stress from external loading. An item with a surface compressive residual stress is, therefore, predisposed to accommodate an imposed external tensile stress of greater value than would otherwise be the case, as postulated in the 'stress shakedown' of the Comet failure analysis.

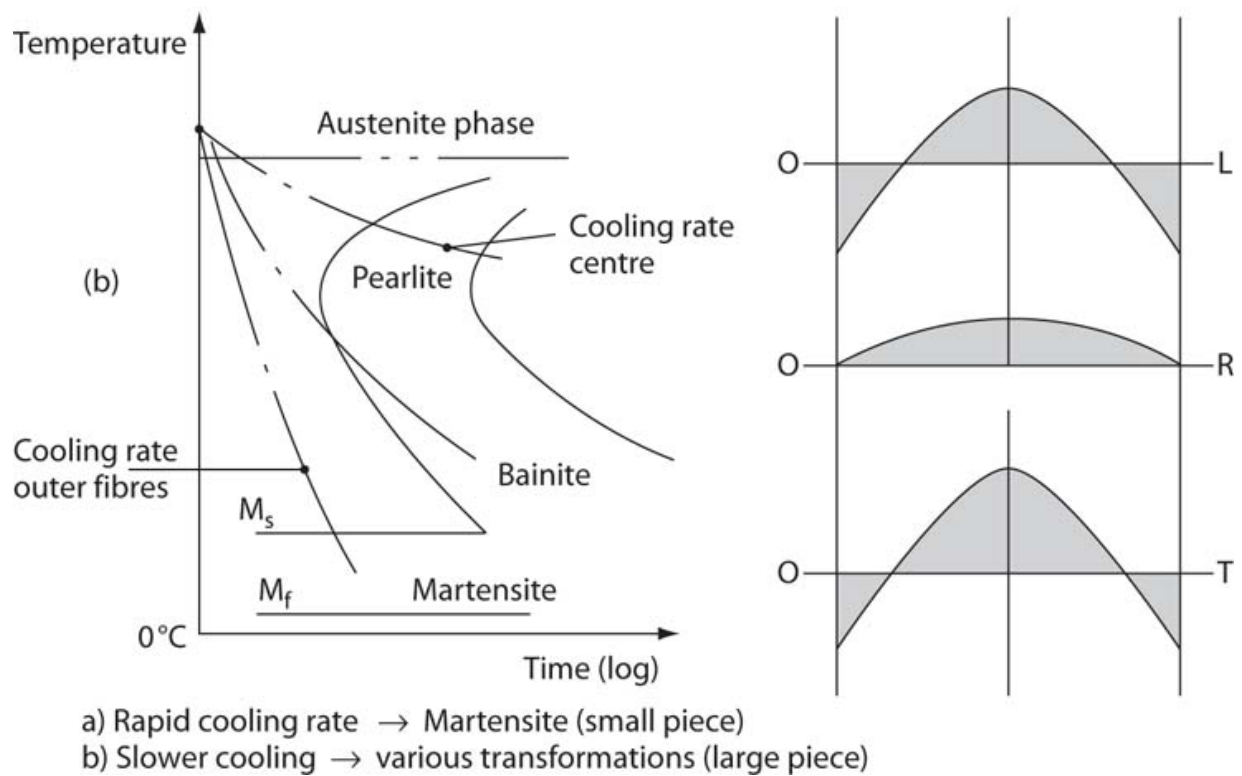
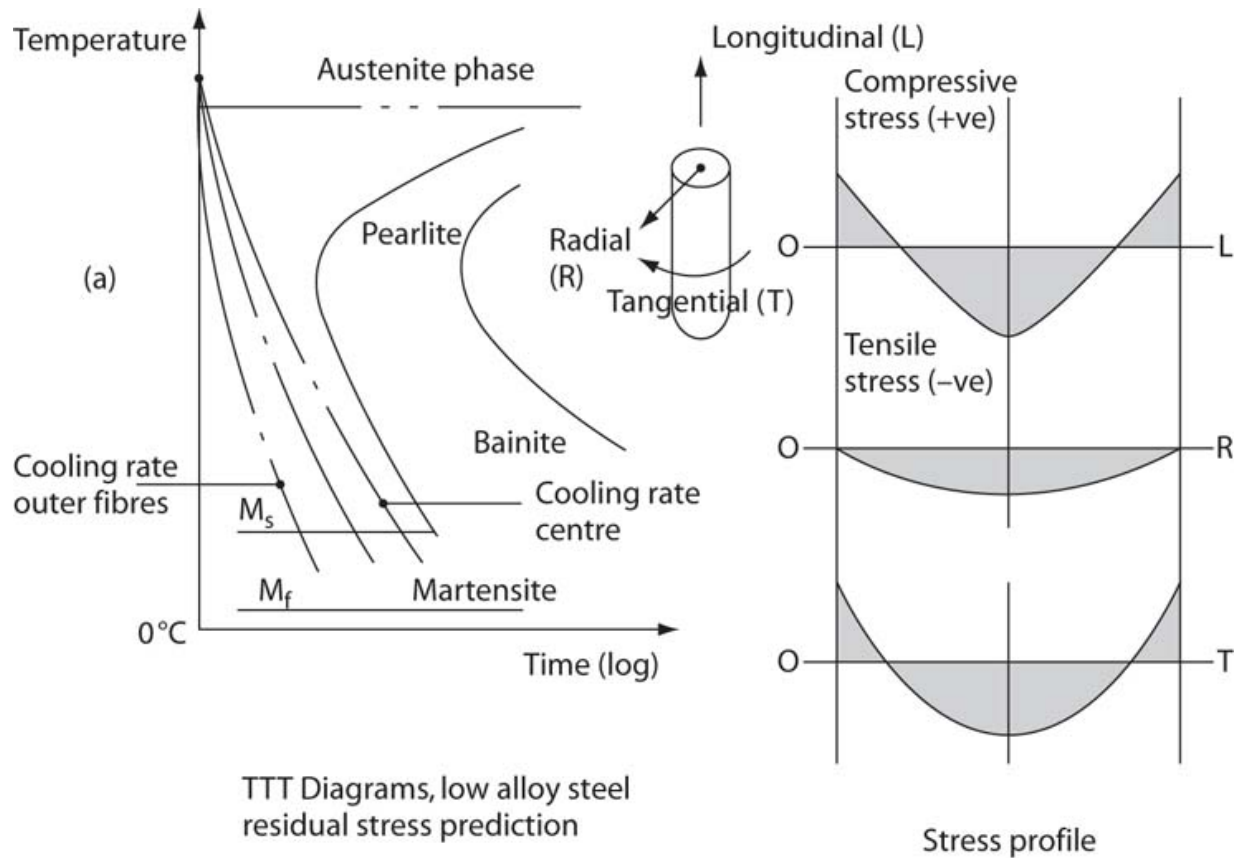


Figure 18 Residual stress on quenching an alloy steel.

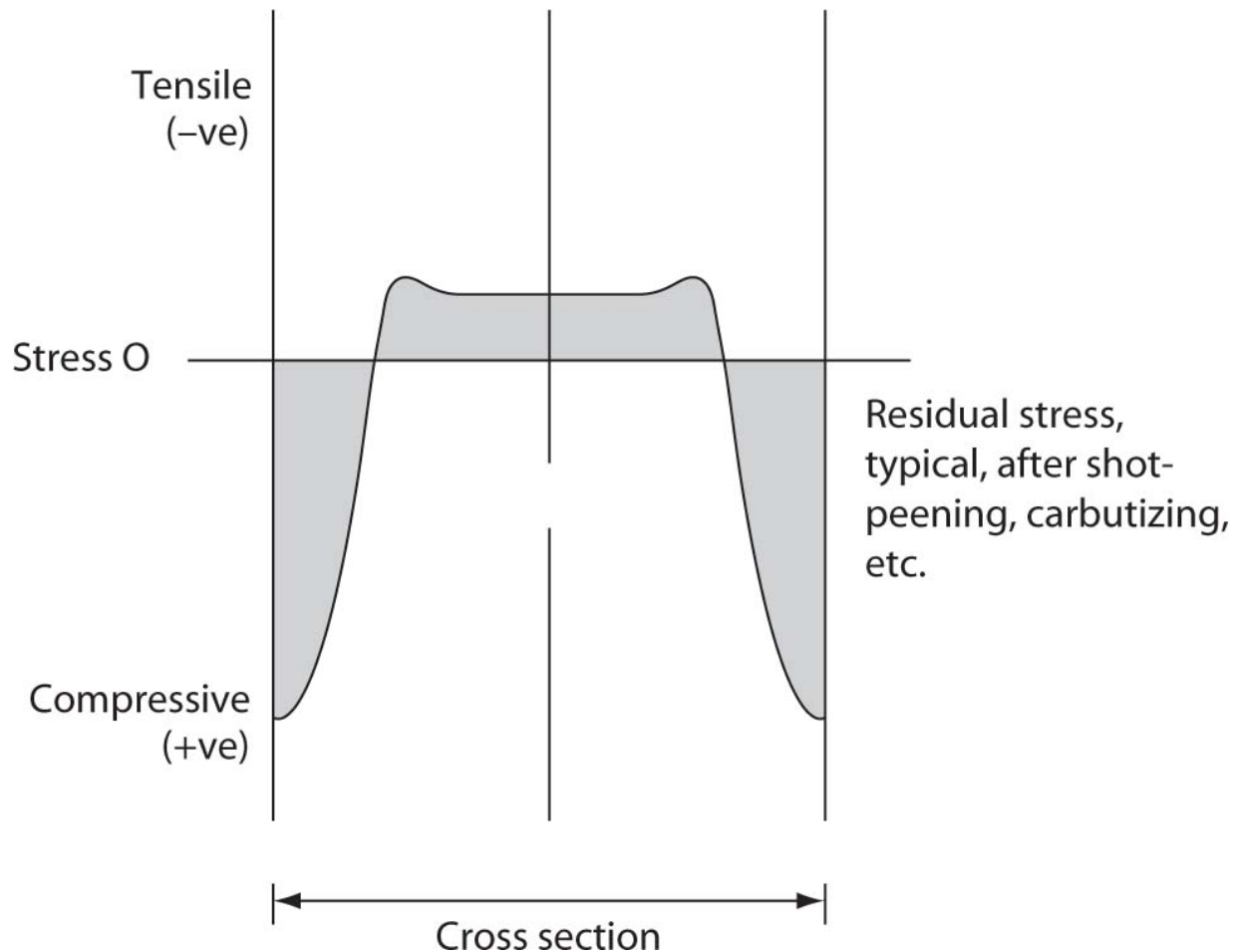


Figure 19 Residual stress after surface treatment.

Conversely, a component already suffering a surface residual stress in tension is in a potentially dangerous condition when further tensile loading is applied. This is particularly the case in fatigue conditions. For components in bending or rotation fatigue, whilst one surface is in tension, the opposite surface is in compression. Reversing every cycle, the residual stresses can be highly damaging and lead to surprising failures.

Even in static situations, the effect of residual stress from machining operations can be very troublesome when trying to obtain very accurate run-outs of precision shafts. Problems can arise with a fine finishing cut, removing a layer of residual stress and causing a small 'bowing' of the shaft. This is not normally a problem with small components, but can be significant in producing large, high-precision shafts for industrial rotating products, like pumps or turbines. It has

even been known for resort to be made to spot-heating, to introduce a residual stress to straighten the shaft. A task for the very experienced, or desperate!

A familiar experience with residual stress is the relaxation that occurs when cutting through a structure, to find it springing apart, sometimes with dramatic effect. This is a common occurrence with welded structures, and indicative of their residual stresses. These normally result in material adjacent to the weld retaining a tensile stress equal to the yield stress, as in [Figure 83](#). This transverse stress is the result of the contraction of the weld (a miniature casting) on cooling, whilst restrained by the bulk of the parent material. It is accompanied by a longitudinal stress of a lower magnitude, as the weld also shrinks in the longitudinal direction. A thin tube, butt welded, will show both the transverse and circular effects of this residual stress, which remains after the plastic deformation has 'pinched in' the tube at the weld, as shown in [Figure 83](#).

The combination of stress concentration, residual stress, and inherent surface flaws all contribute to the potential for early failure of welded components in fatigue conditions, and steps are required to obviate these effects in critical applications.

Combating residual stress

Where the positive virtues of compressive residual stress are not available, then means are sought to alleviate it through thermal or mechanical stress relief, by surface treatment.

For welded structures, 'post-weld heat treatment' (PWHT) is often applied to steel structures and vessels, by slowly heating to $\sim 600^{\circ}\text{C}$ and slow cooling, relaxing the high residual stresses inevitable during fabrication. Whilst beneficial, this generally relieves only ~ 60 per cent of internal stresses; it is also expensive, and can result in distortion of an unsupported structure as the residual stresses are relieved.

Note that the only safe treatment for most stainless steels is a rapid quench from $\sim 1050^{\circ}\text{C}$ in order to retain a fully austenitic structure. This is normally impractical as it can lead to serious plastic deformation at these high temperatures, where the yield stress is low.

In order to avoid the difficulties of high temperature thermal stress relief, a number of alternatives have been successfully developed. These include low temperature stress relief, typically $100\text{--}300^{\circ}\text{C}$. This is often used to allow any dangerous hydrogen build-up, in electro-plated components, to be relieved and so render the internal stresses less critical.

The phenomenon of ‘season cracking’ in some brasses is a manifestation of residual stress arising from forming operations, often seen in pressed cones and shapes for domestic items like door handles and antique furniture (see [Photos 57](#) and [58](#)). This is a stress-corrosion problem and the solution lies in careful material choice and low temperature stress relief. Similarly, some familiar austenitic stainless steels can be susceptible to ‘stress-corrosion cracking’ (SCC). Even without external loading, problems can arise from the internal effects of residual stress, combined with certain environments or contamination. This can occur from chloride in salt spray, or zinc contamination from galvanized mating parts such as nuts or washers.

The most popular alternative to thermal treatments is through mechanical processes to provide a compressive residual stress on the component surface. In theory, this could be achieved with judicious hammering, but in practice this is too imprecise to be trustworthy and is best achieved with professional ‘shot peening’. A specialized technique, it has, when properly applied, been found to have a dramatic effect in applications where fatigue performance is critical. This commonly occurs in rotating shafts, and at the toe of fillet welds, where fatigue crack initiation is always anticipated. Typically depths of shot-peening are 0.05 to 0.25 mm (2–10 thou). (See [Table 29](#).)

The final method considered here is for surface treatment by carburizing, nitriding, or similar. These tend to have the effect shown

in Figure 19, inducing a compressive surface stress. However, they do involve relatively high temperatures and may be difficult to use without distortion to a finish-machined part.

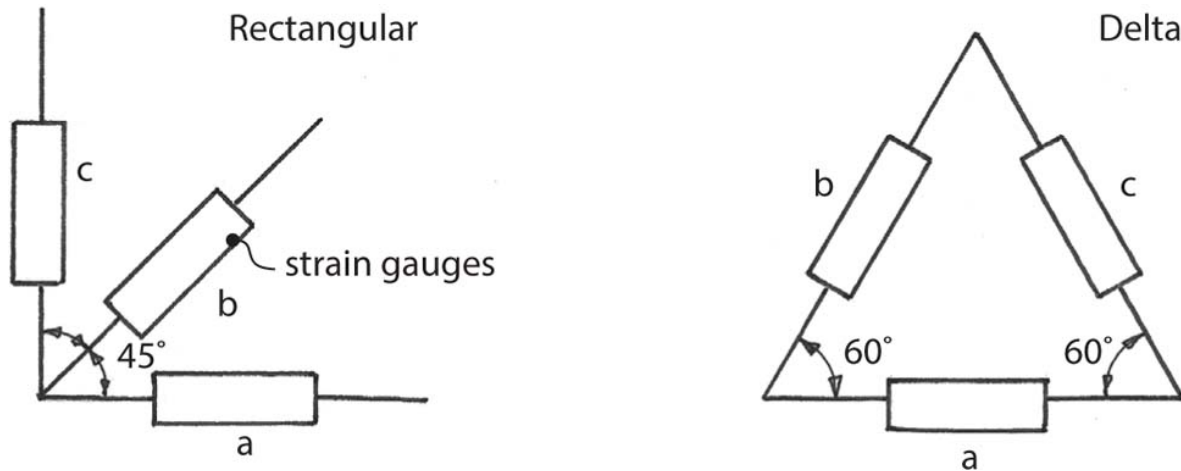


Figure 20 Strain gauge rosettes.

Measurement of residual stress

For the home engineer this is sufficiently difficult as to be of largely academic interest although, as noted before, there are occasions where the effect of residual stress is obvious, such as the warping of a welded structure or springing apart of a fabrication when cut.

Experimental methods have traditionally used photo-elastic models which show stress patterns as loads are applied; strain measurement as the component is systematically machined, so removing layers of stressed material; or X-ray diffraction measurements that follow the microstructural stress from first principles. None of these methods is normally appropriate for our purposes, so a general grasp of the theory, supplemented by rule-of-thumb tests mentioned previously, are the best way forward.

Relatively recent developments in computer modelling have now provided some useful insights into residual stress in real components, but this presently remains the province of industrially funded activities.

Strain measurement

As it is not normally possible to see or measure stress directly, then, provided that only the elastic range of the stress–strain graph is considered, the measurement of strain still provides the best means to determine stress by experimental methods.

However, because strains are usually very small in most practical applications, it is necessary to have accurate measurements. This has led to the long established strain-gauging techniques where sensitive film-type gauges are affixed to the component surface by adhesives. The change in electrical resistance with surface strain is measured with sensitive meters. This enables measurement of the maximum and minimum direct stress ($\sigma_{\max}$, $\sigma_{\min}$), and the orientation (α) of these values with respect to the component's axis.

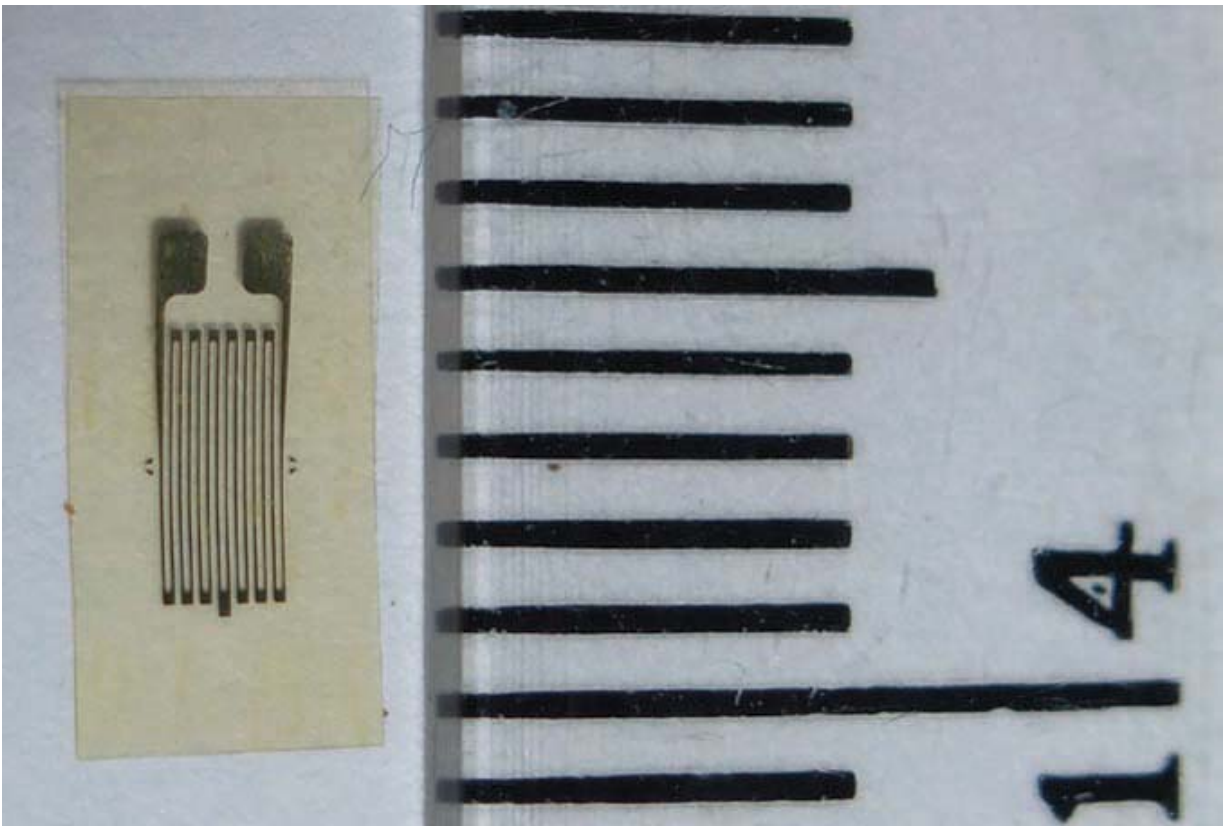


Photo 64 Single resistance strain gauge, ~ 4mm long.

Mohr's Circle

Without recourse to the rather sophisticated theoretical aspects of elasticity (see [Bibliography](#)), we can use the resulting relationships, graphically represented by Mohr's Circle, to make sense of strains as measured by the strain gauges. From these measurements the principal stresses, $\sigma_{\max}$, $\sigma_{\min}$ and α , can then be derived.

In order to apply Mohr's Circle we use well-established practice with strain gauges arranged into 'rosettes' of either rectangular or delta pattern (see [Fig. 20](#)). The rosette is attached to the component at any orientation to a datum line, normally chosen as the main axis of the item. Progressive loading, within the elastic limit, enables a plot of load versus strain at operational loads, which can now be plotted on the Mohr's Circle to determine the principal strains and their orientation to the datum. (See [Figure 22](#).)

Stress evaluation

Whilst it might be tempting to simply obtain stress from our measured strain, as $E = \sigma/\epsilon$, this should be resisted because of the Poisson effect of lateral strain. Whilst construction of the Mohr's Circle provides a useful visual representation of the relationships of the principal strains, it is convenient to employ the result of further elasticity theory (see [Bibliography](#) for derivation) to provide immediate results for $\sigma_{\max}$, $\sigma_{\min}$ and α . Typical values for the material constants, E , G and ν were seen earlier, in [Table 20](#).

STRUCTURAL ANALYSIS

Beam bending

As noted earlier, the correct theory of beam bending stress analysis took a long time to be formulated. Its understanding is vital, as it is soon apparent that loads can be far more readily supported directly in tension or compression, compared to being supported by a beam (see [Figure 21](#)).

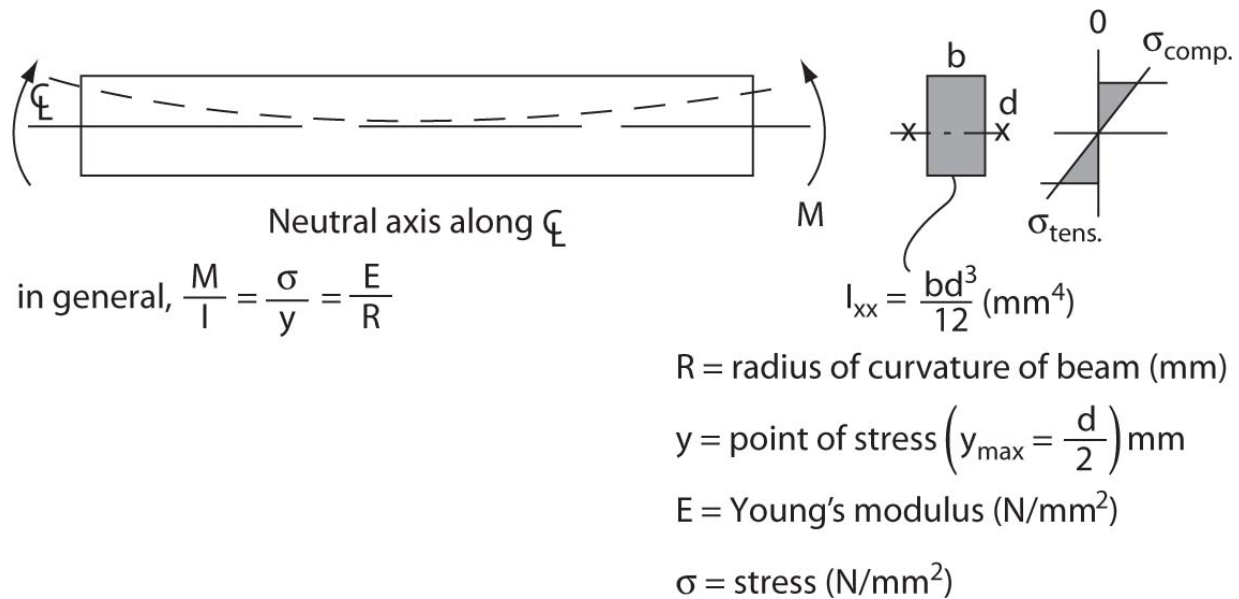


Figure 21 Beam bending (rectangular section)

However, because beams are so fundamentally important to structures and mechanisms, it is valuable to be able to check the stress in any safe loading application. The first important feature concerns stresses across the section. Assuming, for the moment, a symmetrical section about the XX axis, then this loading will cause a maximum compressive stress in the upper fibres, reducing to zero at XX, the 'Neutral Axis' (NA). The lower half mirrors the upper half, but this time as a tensile stress, to a maximum at the outer fibre. Thus, it is intuitively apparent why a beam is less efficient supporting a given load, compared to that in tension. Hence, strong wires are highly efficient load carriers (cycle-wheel spokes, for example). However, in many applications beams are needed to support distributed loads – for everything from bookshelves to bridges.

The relative inefficiency of beams for load carrying can often be used to advantage, such as the controlled deflection of a leaf spring. It is apparent that the behaviour of beams is a function of several parameters, including: cross-section shape; material (σ_y , E); method of support; and load distribution. If we consider a simplified case of a footbridge over a river, not only could it fail by exceeding its tensile strength, but it could also possibly fail by excessive deflection, into the water, whilst still within a safe stress. Perhaps rather less

obvious is failure by lack of stiffness, causing the 'beam' (structure) to twist or distort too far. An example is the Millennium Bridge, which crosses the Thames in London between St Paul's Cathedral and Tate Modern. Enthusiastic pedestrian use soon revealed minor lateral stiffness deficiencies, leading to a gentle swaying motion, cured by extensive use of vibration dampers. Far more extreme cases have been well documented, where slender structures like suspension bridges have suffered from lack of lateral stiffness, allowing excitation by wind loading or traffic flow to set up an excessive vibration. Perhaps the most spectacular example was the Tacoma Narrows suspension bridge in Washington, USA. The large-scale wind-induced vibrations led to the nickname, 'Galloping Gertie'. Often ascribed to a resonant frequency effect, eventual failure was actually by 'aeroelastic flutter'. The deflections increased, from that providing an exciting trip across the bridge to catastrophic failure, fortunately with just enough time for all to safely flee! (But this was not an unknown phenomenon, having previously caused the failure of the Old Chain Pier, Brighton, UK in 1836.)

Whilst the home workshop's projects should be free of concerns about wind and vehicle loading on large structures, nevertheless there are cases where there is a need for appreciation of potential problems, applied on a smaller scale. If a sound calculation cannot be effected, then a fail-safe means is required to accommodate a potential failure, if proof testing is not possible.

MOHR'S CIRCLE

The construction in this example is detailed and is surprisingly straightforward by this method.

$$\epsilon_A = 300 \mu\epsilon$$

$$\epsilon_B = -120 \mu\epsilon$$

$$\epsilon_C = 200 \mu\epsilon$$

1. Vertical lines indicate strain magnitude. Begin with **Datum line**.
2. Draw verticals at scale distances of: 300 for A, 200 for C, and -120 for B, all from Datum line.
3. Choose an arbitrary point in the central line (C here), and sketch rosette at point O. Follow rosette line for A (horizontal) to meet line A at D. Follow rosette line for B (45°) to

meet B at E.

4. Draw circle through E, O and A by bisecting line O–D, bisecting line E–O; where the bisecting lines meet is the circle centre, F.
5. Draw circle as above. The centre line H–G provides max and min strains: E max = 603μ, Emin = –125μ, as measured from Datum line.
6. Angle of εmax from line O to G, measured from line O–D, α ~ 41.7°, gives orientation of max strain. Min strain at 90° to max, line O–H.

Check of Mohr's Circle for principal strains

Check against calculated method (ref. *Mechanical Metallurgy*, Dieter, E.D., 1961)

From the original data here:

$$\epsilon_A = 300 \mu\epsilon$$

$$\epsilon_B = -120 \mu\epsilon$$

$$\epsilon_C = 200 \mu\epsilon$$

$$E = 30 \times 10^6 \text{ psi}$$

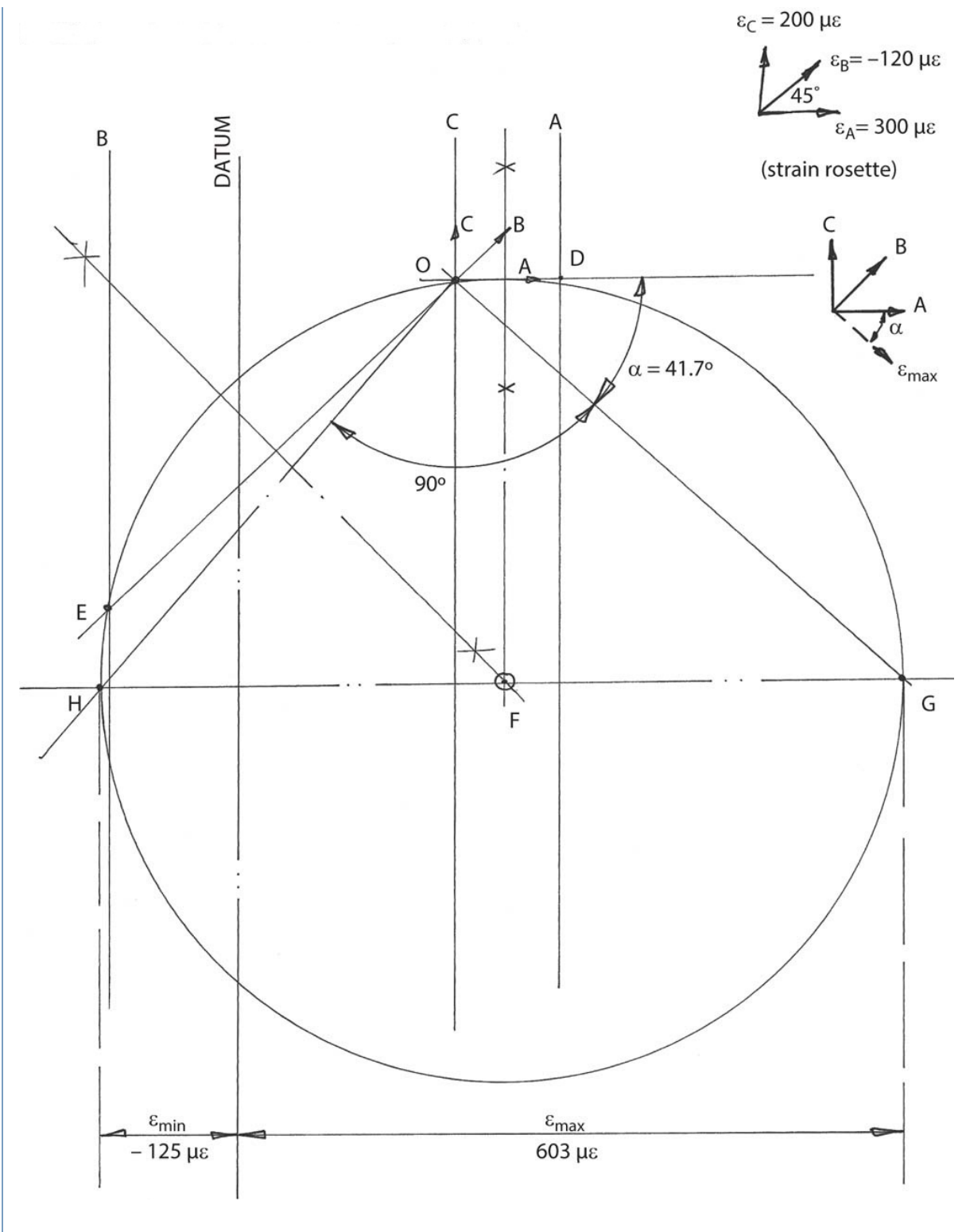
$$\nu = 0.3$$

For a rectangular rosette: see equation below.

Comparing to Mohr's circle drawing method, the value of 18,642psi, then the error of my approximate drawing is –3%, which is acceptable for normal practice.

$\sigma_{\max} = \frac{E}{(1-\nu^2)} (\epsilon_{\max} - \nu \epsilon_{\min}) = 18,642 \text{ psi}$ ~18.6 ksi $\sigma_{\min} = \frac{E}{(1-\nu^2)} (\epsilon_{\min} - \nu \epsilon_{\max}) = 1,842 \text{ psi}$ ~1.8 ksi for a rectangular rosette: $\sigma_{\max} = \frac{E}{2} \left[\left(\frac{\epsilon_A + \epsilon_C}{1-\nu} \right) + \left(\frac{1}{1+\nu} \right) \sqrt{(\epsilon_A - \epsilon_C)^2 + [2\epsilon_B - (\epsilon_A + \epsilon_C)]^2} \right]$ $\sigma_{\max} = \frac{30 \times 10^6}{2} \left[\frac{(300 + 200) \cdot 10^{-6}}{(1-0.3)} + \frac{1}{(1+0.3)} \sqrt{(300 \times 10^{-6} - 200 \times 10^{-6})^2 + 2 \times (-120 \times 10^{-6}) - (500 \times 10^{-6})^2} \right]$	$\text{for: } \nu = 0.3$ $E = 30 \times 10^6 \text{ psi}$ $\epsilon_{\max} = 630 \mu\epsilon$ $\epsilon_{\min} = 125 \mu\epsilon$
$\sigma_{\max} = 19,268 \text{ psi} \sim 19.3 \text{ ksi} \quad \text{--- calculated}$ $(\sigma_{\max} = 18,642 \text{ psi} \sim 18.6 \text{ ksi} \quad \text{--- drawing})$	

Figure 22 Construction of Mohr's Circle for principal strains.



Simple bending stress

The first assumption is a stress range within the elastic limit. Without troubling about derivations (see [Bibliography](#)), for the simple beam we have $\sigma = My/I$, where stress is in N./mm², M is the bending moment in N./mm; y the distance from neutral axis to point of concern, in mm; and I is the 'second moment of area' in mm⁴. We also have the useful relationship, of $M/I = \sigma/y = E/R$; where E (Young's Modulus) is in N./mm²; and R, the radius of beam curvature, is in mm, as shown in [Figure 21](#).

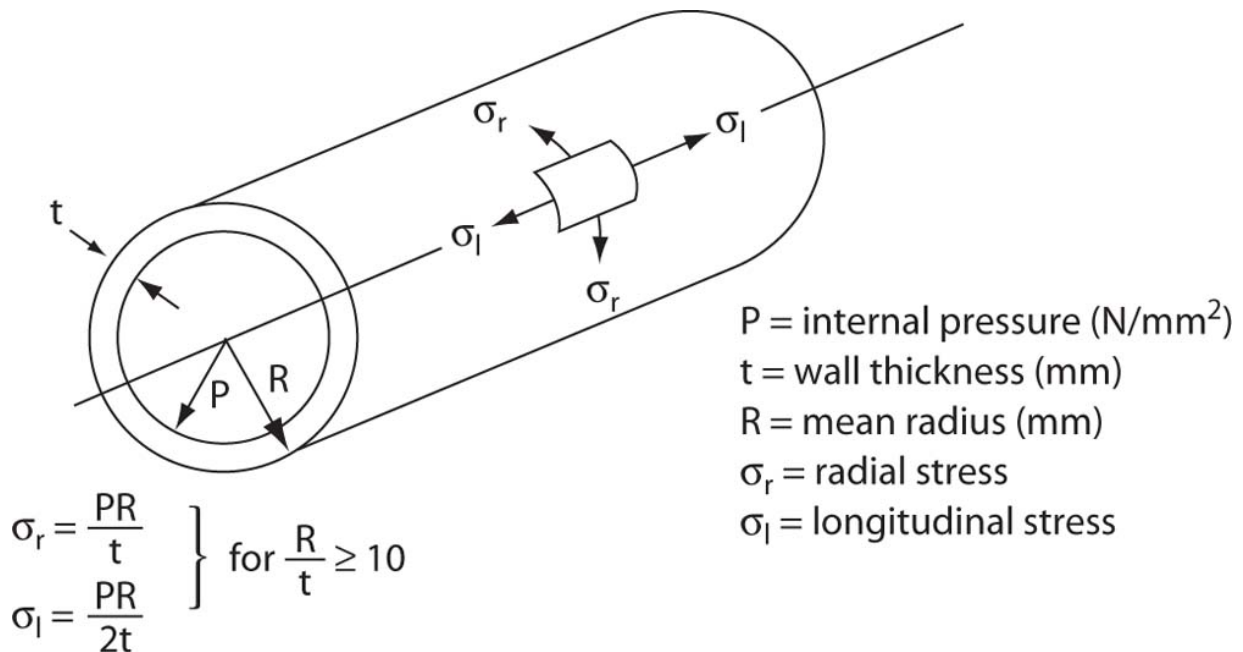


Figure 23 Thin cylinder stress under internal pressure.

Second moment of area (I) and bending moment (M)

This is a slightly unusual concept, in units of mm⁴, and best tackled through some simple examples, shown in [Figure 24](#). Basically, it is a shape factor which enables a precise calculation to be made to represent the effect of the beam cross-section in the above formulae. For regular sections such as rectangles or round bar, it is found from the simple formulas of [Figure 24](#), but can be calculated (or found by graphical means) for any shape. Fortunately, all but the most unusual shapes are readily found in reference data books (see [Bibliography](#)).

‘Bending moment’ (M) is the bending effect induced by the applied load(s), and can be determined, as shown in [Figure 27](#), by drawing up a ‘bending moment diagram’, if required. Normally, we are only interested in maximum stress and deflections in beams, and these can be found by short-cutting to the formulae as above.

In principle, bending moment is the product of load times distance ($M = P \times L$), from the datum at which it is applied.

Beam support and loading

The way a beam reacts to a load is, clearly, dependent on the support method. In order to provide ready answers to practical problems, we simply need to employ the appropriate model, as tabulated in the Data Books.

The models shown are normally sufficiently near to practical applications, but it is a simple matter to use one that gives a conservative result, higher stress, for safety. For instance, a ‘simply supported’ beam is clearly inferior to the ‘built-in’ case for load carrying. A bolted frame may be assumed to act like a pin-jointed structure, if only a single small bolt is used at the supports, whereas a fully welded beam, welded to a ‘massive’ support, can be considered ‘built-in’.

The way in which a load is applied to the beam also influences the beam’s response. These loads are modelled as either ‘point loads’ or ‘uniformly distributed loads’ (UDL). They can co-exist along a beam, and the models provide examples from which to choose. Whilst structures, such as flooring, are often designed using the UDL models, there are many devices where we encounter the more onerous application of point loads, such as in the leaf spring.

Application

An understanding of beam bending theory is particularly useful where we seek to optimize or modify a structure, to save weight or material, whilst still safely retaining function. It can be used in

conjunction with a range of materials and component sections in order to select the best design. A series of iterations can be tried to see how the various factors, such as material stiffness, can be judged against available component sections, to provide the best choice based on performance, strength, weight, cost and availability. The success of the British Olympic cycling team of 2012 again demonstrated the contribution of this, in an apparently simple and mature product, which can still be improved by the dedicated application of this science, albeit at an extreme level of activity.

Pressure stresses in cylinders

The case of internal pressure in cylinders and spheres is a common feature of products from the home workshop, including familiar items like steam boilers. Less common is external pressure applied to vessels. Whilst this is vitally important in industrial applications from vacuum vessels to sub-sea craft, we concentrate on internal pressure. Should low (below atmospheric) pressure be a consideration, then the ASME VIII and PD 5500 publications provide full analytical details, as well as in all other aspects of serious pressure vessel design (see [Bibliography](#)).

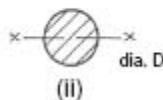
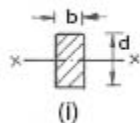
Second Moment of Area, in general
about axis X - X, I_{xx} is :-

$$I_{xx} = \int_0^{y_i} y^2 dA \text{ (mm}^4\text{)}$$

$$\text{and Centroid (centre of area)} = \int_0^{y_i} y dA \text{ (mm)}$$



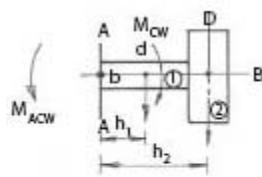
Fortunately, Second Moments of Area and Centroids can be found for various shapes as the [Bibliography](#), but for rectangles and rounds we have:-



$$(i) I_{xx} = \frac{bd^3}{12} \text{ (mm}^4\text{)}$$

$$(ii) I_{xx} = \frac{\pi D^4}{64} \text{ (mm}^4\text{)}$$

Centroids can be found for simple shapes by using moments about a point, ie. $M_{cw} = MACW$.
For 'Tee' section take moments about A-A :-

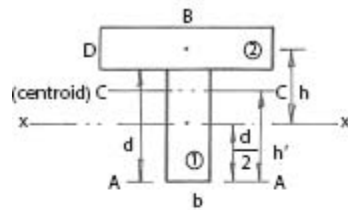


$$\begin{aligned} \text{Area } A_1 &= b \times d \\ M_{CW} &= M_{ACW} \\ (A_1 + A_2) h' &= (A_1 \times h_1) + (A_2 \times h_2) \\ \text{ie. } (32 + 20) h' &= (32 \times 4) + (20 \times 9) \\ 52 h' &= 128 + 180 \\ h' &= \frac{308}{52} \sim 5.923 \text{ mm} \end{aligned}$$

$$\begin{aligned} b &= 4 \text{ mm} \\ d &= 8 \text{ mm} \\ D &= 2 \text{ mm} \\ B &= 10 \text{ mm} \\ h' &= ? \\ A_1 &= b \times d \\ &= 32 \text{ mm}^2 \\ A_2 &= B \times D \\ &= 20 \text{ mm}^2 \\ h_1 &= 4 \text{ mm} \end{aligned}$$

So, centroid is 5.923 mm from A-A ($h' = 5.923 \text{ mm}$)

Second moment of area and centroid can be found thus.



As before, now I_{xx} has
for A_1 , $h = 0 \text{ mm}$
for A_2 , $h = 5 \text{ mm}$

$$\begin{aligned} I_{xx} &= I_{1,xx} + [I_{2,xx} + Ah^2] \\ &= \frac{bd^3}{12} + \left[\frac{BD^3}{12} + (B \times D \times h^2) \right] \\ &= \frac{4 \times (8)^3}{12} + \left[\frac{10 \times (2)^3}{12} + (10 \times 2 \times (5)^2) \right] \\ &= \frac{4 \times 512}{12} + \left[\left(\frac{10 \times 8}{12} \right) + (20 \times 25) \right] \\ &= \frac{2048}{12} + \frac{80}{12} + 500 \\ I_{xx} &= 677.3 \text{ mm}^4 \end{aligned}$$

This provides the Second Moment of Area about the axis X-X, drawn through the centre of section ①. However, in bending, the beam will have its *Neutral Axis* through the centroid, axis C-C. Therefore, maximum stress is measured from the centroid to the fibre furthest from the centroid, which is 'y' in the formula, $\sigma = \frac{My}{I}$; in this case, $y_{\max} = 5.923 \text{ mm}$.

Figure 24 Second moment of area and centroid.

Internal pressure

A simple analysis provides guidance, provided we consider the thin-walled vessel, where diameter is greater than ten times wall

thickness ($D/t > 10$). Whilst the result can be derived from first principles (see [Bibliography](#)), we adopt the stress based on an element which is subjected to the principal stresses as shown in [Figure 23](#). As this is a thin wall, it produces simple biaxial stress, neglecting the through-thickness stress that needs to be accounted for in thicker vessels.

From $\sigma_r = PR/t$ and $\sigma_l = PR/2t$, it is clear why pressure failures normally follow a longitudinal split (cf. Comet failure described earlier in this chapter), as shown in [Figure 26](#). For a sphere, the principal stresses are equal at $\sigma_r = PR/2t$.

Thus, the most efficient vessel is a sphere, or cylinder capped with hemi-sphere. Where we have to utilize flat ends or rectangular sections, the stress rises significantly and analysis becomes more complex. In practice, capping off a nozzle with a flat plate requires plate of far greater thickness than the vessel, i.e. $\sigma_r = 2PR^2/3t^2$ (see [Figure 25](#)).

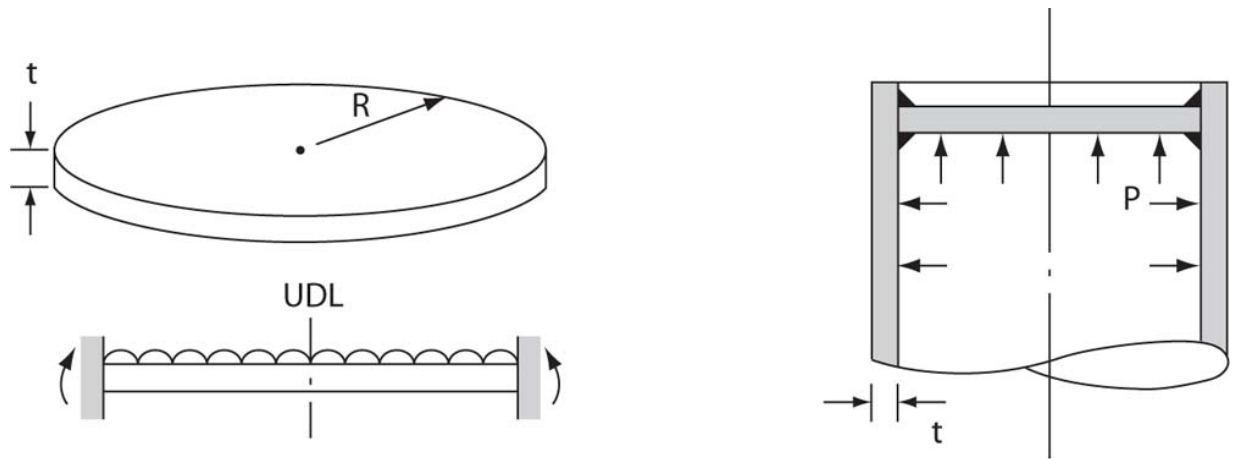
Where simple cylinders and spheres cannot be employed, as in traditional model locomotive boiler designs, the inferior pressure-retaining characteristics of flat sections have to be compensated-for by restraining struts or ribs, otherwise a variety of material thicknesses would be required.

Expansion effects

A change of temperature will produce a change of length (ΔL) in metals according to its characteristic expansion coefficient (α), length (L) and temperature range (ΔT), as $\Delta L = L.\alpha.\Delta T$. This produces a strain of $\Delta L/L$, and thus stress (σ) = $E. \Delta L/L$, if the length change is restrained. This was the principle of the old 'clackity-clack' of railway lines laid with gaps for expansion and contraction with the seasons. With the gaps replaced by welded joints and the continuous lines pre-tensioned, the ensuing strains are accommodated as safe internal (residual) stresses.

For a small unrestrained boiler, the expansion is inconsequential, as shown in [Table 21](#), but where lengths are significant then

movement can be noticed. For a spectacular example, the Concorde grew by some 12 inches (300 mm) during supersonic flight, compared to its temperature on the ground, due to the frictional heating at such great speeds, faster than a bullet! Wherever there are materials of different coefficients joined together, or parts are restrained, it is worth checking this effect. Note that, compared to mild steel, stainless and aluminium have higher expansion rates by ~50 and 100 per cent respectively. A mild steel bar 1m long, heated from 0° to 100°C, will expand ~1.2mm ($L = 1000 \times 100 \times 12 \times 10^{-6} = 1.2 \text{ mm}$).



Stress on flat plate
under pressure, welded in
(model, uniformly distributed load - UDL and built in edge)

$$\text{Stress in plate } \sigma_p = \frac{PR^2}{t^2} \times 0.67 \text{ N/mm}^2$$

$$\text{Compare to stress in tube wall } \sigma_r = \frac{PR}{t} \text{ N/mm}^2$$

Figure 25 Stress in flat plate under pressure.

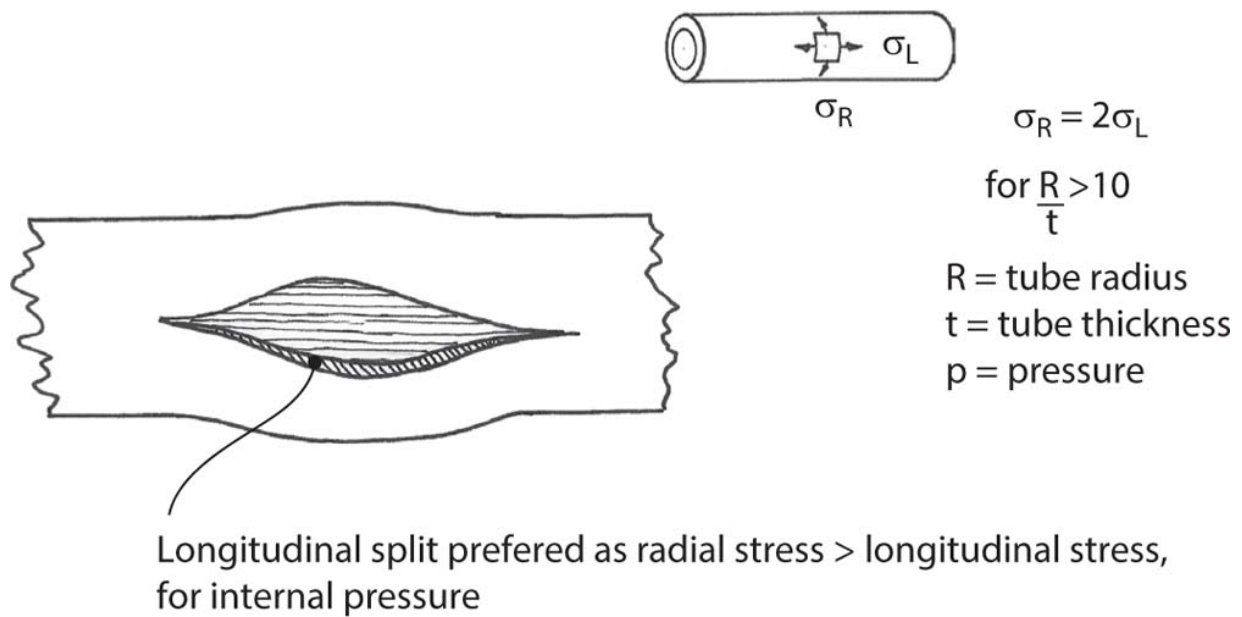


Figure 26 Tube failure under internal pressure.

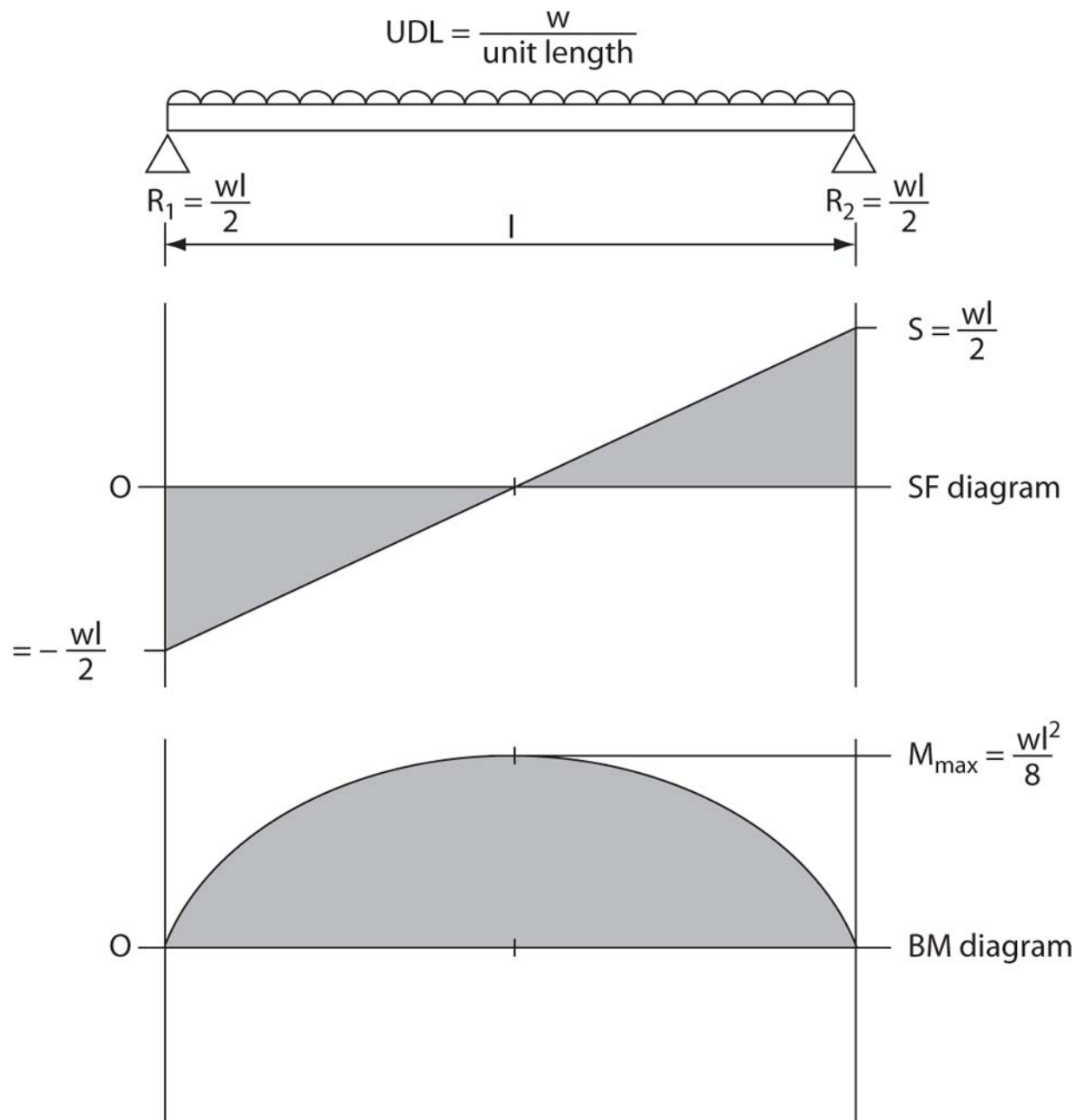


Figure 27 Simply supported beam with UDL, shear force and bending moment diagrams.

Table 21 Coefficient of expansion (α) (expansion, $\Delta = L \alpha \Delta T$).

Material	α ($\times 10^{-6}$)
aluminium	23
brass	19

copper	17
nickel	13
carbon steel	11
stainless steel	18
tungsten	4.5
silicon	3
diamond	1

L = length (mm), Δ = change in length (expansion, mm), ΔT = temperature change ($^{\circ}\text{C}$), coefficient of thermal expansion α (unit-less)

Another impact is the effect on measuring equipment where accuracy is paramount, so it is always good practice to keep reference standards (such as engineers' slips), micrometers and work pieces at consistent temperature.

Cases arise where expansion is significant and thermal stresses need to be avoided, as in vessels with heater tubes fixed into tubesheets with joints at both vessel ends, the 'fixed tubesheet' design. Typically, the tubes and vessel are from different materials like stainless and mild steel, and heating will induce thermal stress. The usual solution is to introduce a 'bellows' convolution into the vessel, which retains the pressure but allows stress-free lateral movement. A specialist area, thin and thick wall bellows design requires reference to experienced suppliers or substantial design expertise.

Shear and torsion loading

Shear

This is an important type of loading and follows on from that of direct tensile or compressive loading. In shear, the material sustains a stress equal to the applied load divided by the resisting section but, unlike direct tension where the resisting area is normal to the load, in shear the resisting section is aligned with the load, as shown in [Figure 28](#). It can also provide a method of separating (shearing) material, to provide failure at lower loads than in tension, and is the mechanism operating in the lathe cutting tool and guillotine.

In the coupling of rods to levers, use is often made of clevis pins and yokes, The initial calculation of these stresses is straightforward, but if clearances increase with wear then the mode of failure can introduce an element of bending. Note that these components are often balanced, presenting two areas of shear, so each 'shear face' sees only half the total shear stress.

Mechanical connections, like bolts and rivets, may be normally subjected to shear loads, as in [Figure 29](#), and a simple calculation provides the shear stress. In practice, there is normally a tensile load applied by the bolt to the plates. This produces a frictional force resisting shear, but care should be taken as wear or corrosion can reduce this, leaving the bolt to support the load in simple shear.

Welded structures, by their continuous load path, always have a shear component that must be considered, as described later in 'Welded Joints'.

In beam design, besides the primary effect of moments and bending with the Bending Moment Diagram, we have an analogous situation for shear. Thus, a 'Shear Force Diagram' can be constructed to illustrate the values of shear force that apply at all points along the beam, as demonstrated in [Figure 27](#). Thus, the Shear Force Diagram provides an opportunity to check that the beam will safely support the applied loads.

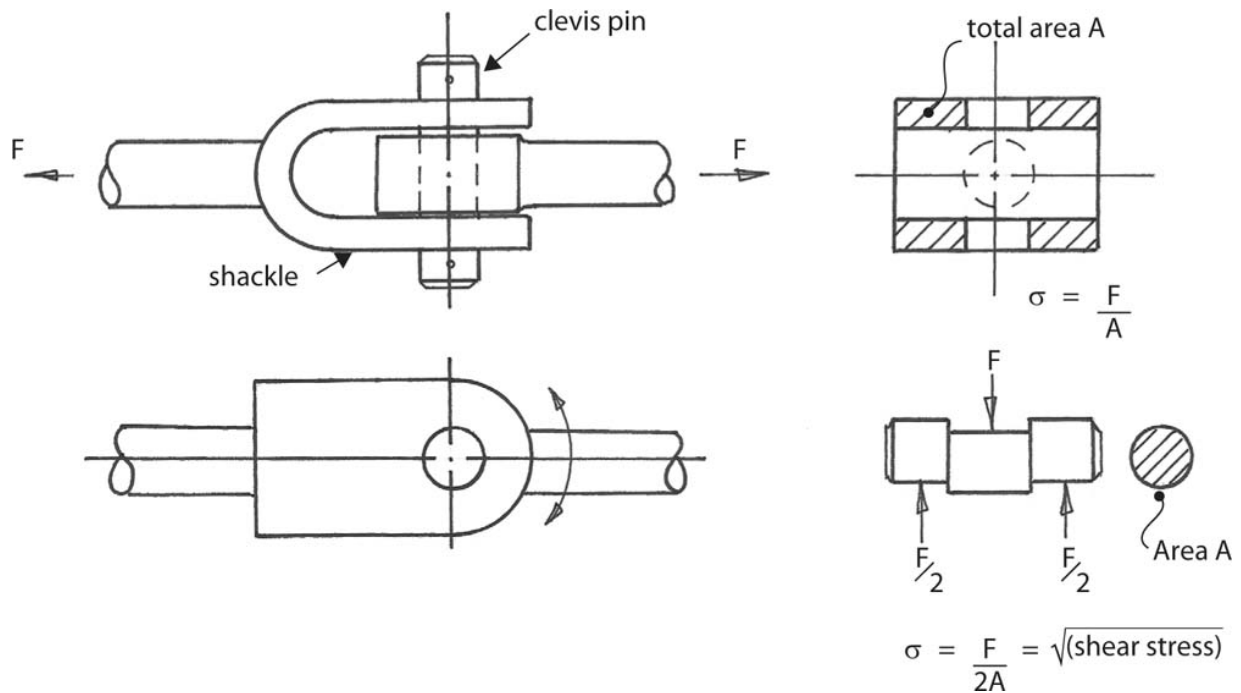


Figure 28 Clevis pin and shackle.

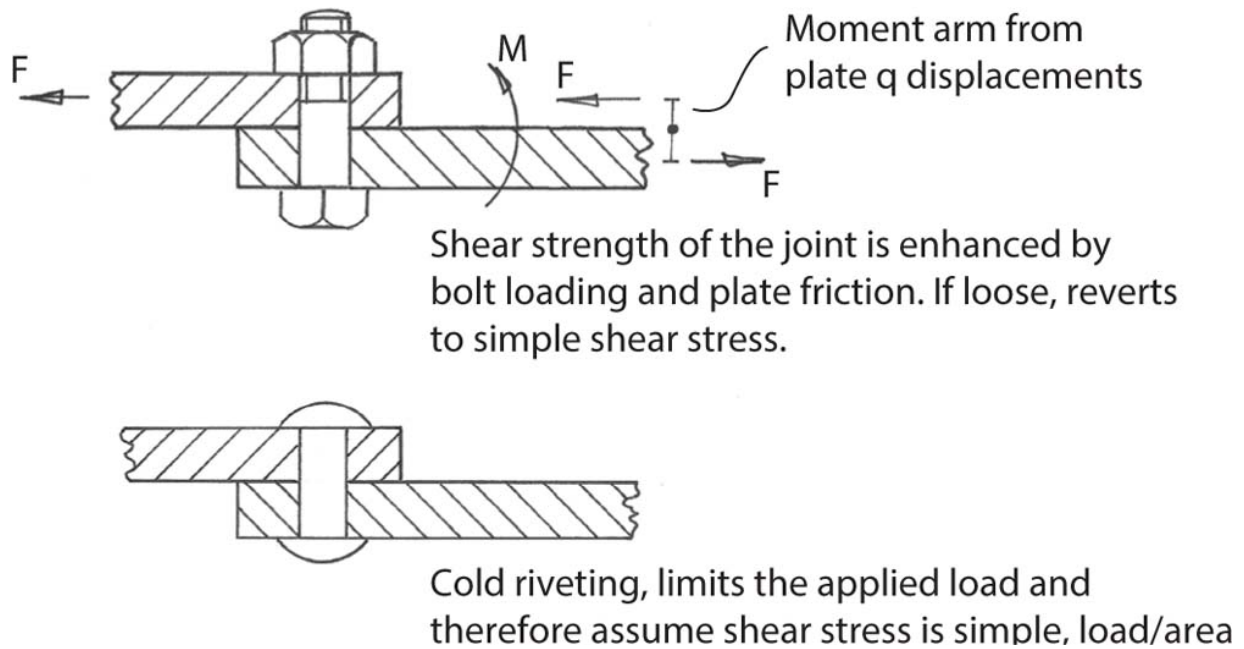


Figure 29 Bolted and riveted joints.

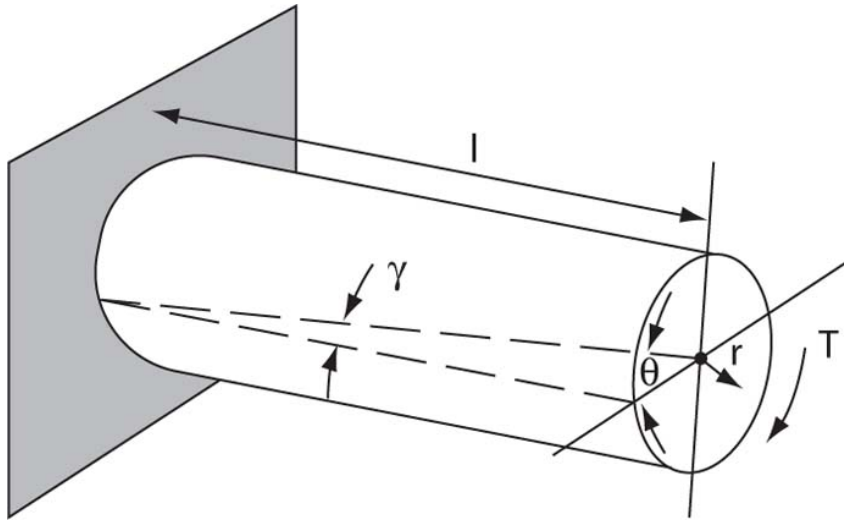
Torsion

This is the final fundamental style of loading we consider, applying to shafts that transmit power and an important consideration in many mechanisms and structures. Evaluation of stress due to torsion is achieved by describing the applied force, multiplied by the 'moment arm' (see [Figure 30](#)). This loading factor is then applied to a material section factor (J). The 'Polar Second Moment of Area' (J) is analogous to the 'I' parameter of beam-bending. It can be incorporated into the very useful formula, with the material modulus (G) and point of stress (r), to obtain the relationship: $T/J = \tau/r = G/\theta$. This equation follows a similar pattern to beam bending, relating applied load to supports, section parameters, material elasticity, to stress and deflections, $M/I = \sigma/y = E/R$.

Another particularly important aspect of torsion concerns coil springs or torsion bars, in spring design. A specialized subject, it is nevertheless accessible to the home engineer (see [Bibliography](#)). Spring design and manufacture can be useful where it is difficult or impossible to obtain an original, and would otherwise render the project inoperable. Spring manufacture often requires a satisfying combination of mechanical design to provide the material dimensions (wire diameter, coil diameter, number of turns and length), and material characteristics. Apart from small and lightly-loaded springs, this is likely to require judicious material selection, heat treated to enable coiling, then final heat treatment to restore the strength required for service conditions, including fatigue resistance. Such an exercise makes one wonder at the performance and reliability of the ubiquitous spring, so integral to modern life (see [Chapter 5](#)).

Torsion also provides a means of evaluating power, from the relationship described above between torque (T) and shear stress (τ), where strain measurement can be used to find the shear stress. Limits to torque can be simply achieved by the use of 'shear pins' or keys, in shafts. The pin is calculated to shear before the torque rises beyond the prescribed limit. This can be a useful feature in machinery where a powerful motor could cause damage to machine or person if something goes awry, such as a tool jamming, or a lawn

mower cutting-cylinder encountering something with a shear strength greater than grass!



In general, $\frac{T}{J} = \frac{t}{r} = \frac{g\theta}{l}$

T = torque (rotation moment atm N.mm)

J = 2nd moment of area in torsion mm^4

θ = angular rotation (at end) radius

g = modulus of rigidity N/mm^2

t = shear stress N/mm^2

γ = angular rotation (at length) radius

r = point on radius (max stress at r_{max})

$\alpha = t/\gamma$

Figure 30 Bar in torsion.

Welded joints

There is clearly a need, in a rational design, for welds to be sized. In practice, it is remarkable how often one sees an engineering drawing with a simple description, 'continuous fillet weld' (CFW), without specifying the weld size. This is surely no less excusable than if a similar bolted joint was specified as merely 'bolt here' – unthinkable!

However, safe weld-sizing may be fraught with difficulties and requires careful consideration, even by the very experienced. In fact,

experience leads one to be wary of over-confidence in this specialized field.

There are several considerations particular to welded joints that need to be examined, beginning with some of the basic assumptions previously used. For instance, stress is assumed to flow in a uniform manner, and mechanical properties are approximately the same in all directions. In welds, with the exception of carefully flushed-off butt joints, a discontinuity – and thus stress raiser – is formed. Welds are never perfectly uniform; some perfectly sound welds can appear remarkably ‘knobbly’. By contrast, an apparently smooth and uniform weld could suffer from a disastrous lack of fusion, rendering it useless. As noted earlier, restrained welds normally produce a residual stress of yield stress magnitude, in the adjacent material. If unrestrained, they can produce significant distortion.

The weldment

Even a well-made weld will, by its very nature as a small casting introduced between the materials to be joined, provide a metallurgical discontinuity, and a preferred site for corrosion in critical applications. If a more ductile weldment is designated, this can overcome some of these undesirable features, although the ultimate strength may suffer.

Despite the above litany of potential difficulties, fusion welding offers some spectacular and unique means of joining metals, or reinstating material lost through over-machining, or environmental degradation, such as corrosion. It may also present a means for protecting the base metal from corrosion or wear, by overlaying methods. It is also, generally, a fast and efficient method of joining, indeed so useful that a welding facility can often be as much a part of the home workshop as the small lathe.

However, whilst metal removal by the processes described in [Chapter 2](#) tends to make a relatively small impact on the workpiece, welding has all of the power and pitfalls that apply to the heat treatment processes described in [Chapter 5](#). Therefore, sizing of

welds needs a knowledge of the material in hand, and the following is aimed primarily at design in mild steels. Whilst a wide spectrum of materials is weldable, including very high integrity work with the low alloy steels, they do demand an understanding of the metal system, as outlined in [Chapter 5](#).

NON-DESTRUCTIVE EXAMINATION (NDE)

The integrity of the weld, touched upon above, is another critical factor. A whole area of activity, 'non-destructive examination' (NDE), has grown up to address this. NDE is principally concerned with the evaluation of welds, and contrasts with the destructive method noted in [Chapter 4](#). Whilst the latter is invaluable in revealing weld profiles and internal or hidden defects, it does (obviously) render the item useless, so that only samples and not the actual product can be tested.

The first, most important and often overlooked method in NDE is the visual examination – as my good friend Ron often says, 'Use these!' – pointing with gusto to his 'mince pies'. Visual examination can be used to pick up most surface defects, such as porosity, lack of penetration, and solidification cracking, as well as assessing the general appearance. The profile of a weld can be very misleading, and experience is required to effect a conclusive evaluation, for example in multi-run 'manual metal arc' ('stick') welds of optimum quality, which can often appear less attractive than a poorer weld that appears smooth, owing to the inherent characteristics of the welding consumable. For high-quality work, 'basic' coated electrodes usually provide a 'knobbly' profile, compared to that of a 'rutile' electrode. However, the former is designed to generate the lowest hydrogen during welding, and the latter can be a source of hydrogen cracking, very important in critical work involving higher strength or restrained joints.

Surface examination can be further enhanced by methods such as 'dye penetrant inspection' (DPI) and 'magnetic particle inspection' (MPI), the former being a process practical for amateur use, requiring only aerosol cans of penetrant and developer and an established technique (see [Bibliography](#)).

A more exacting NDE evaluation requires volumetric examination, definitely a specialist operation but available as a service from the many small NDE businesses. The principal methods are radiography (RT) and ultrasonics (UT), whilst many others exist for industrial purposes. RT uses either a gamma radioactive material source, such as thorium, selenium or ytterbium. These sources are contained in safe vessels and exposed to the job with a film record in a manner similar to medical RTs. X-ray radiography is considered more exacting, but uses an X-ray generator to provide the source and thus is limited for access in some applications.

RT provides a hard copy negative film that can be pored over in the search for internal defects. Interpretation of RT films requires considerable experience, but relies on the same principles as medical radiography. It provides only a twodimensional picture and multiple shots may therefore be required to pin-point a defect.

UT has traditionally relied upon a manual control of a probe over the surface, with output on an oscilloscope, requiring experience in its interpretation and without a

hard copy. It is a method sensitive to picking up linear defects like cracks and 'lack of sidewall fusion', but relies heavily upon the skill and experience of the UT operator.

All volumetric NDE can be a 'double-edged sword' as repairs of marginal defects can cause far more damage, such as residual stress, than 'leaving well alone'. Sometimes records can get disrupted and many miles of underground and sub-sea pipelines have been scrapped on such doubts. Conversely, large product containing buried large and unseen hydrogen cracks has resulted in catastrophic failures of pressure vessels and lifting equipment (see [Chapter 4](#)). Evaluation of welds in critical applications demands a range of techniques and experienced interpretation, unless suitable precautions are designed-in for fail-safe operation.



Photo 65 Locomotive boiler repair, visual examination in natural light (and cold!).

Butt welds

There are two fundamental types of weld joint, the butt and fillet. Butt welds are joints between the edges of two pieces of material, as shown in [Figure 82](#) in [Chapter 5](#). The typical appearance shows a weld cap, usually and erroneously referred to as 'reinforcement'. The ideal profile is a smooth surface which provides no discontinuities or stress raisers. A properly designated and produced weld will have no difficulty matching the base metal strength, and the concern is principally for the region adjacent to the fusion line, the 'heat affected zone' (HAZ), which normally concerns material a few mm into the parent metal. Fortunately, as noted earlier, mild steel is relatively tolerant to HAZ damage.

Providing that the butt weld is free from significant defects, as 'non-destructive examination', has full fusion, and is of the correct designation (referred to as the 'welding procedure'), then the joint

can be ignored in a simple stress analysis. A sound weld can be regarded as 'integral' to the material and disregarded unless particular conditions, such as fatigue or corrosion, are applied.

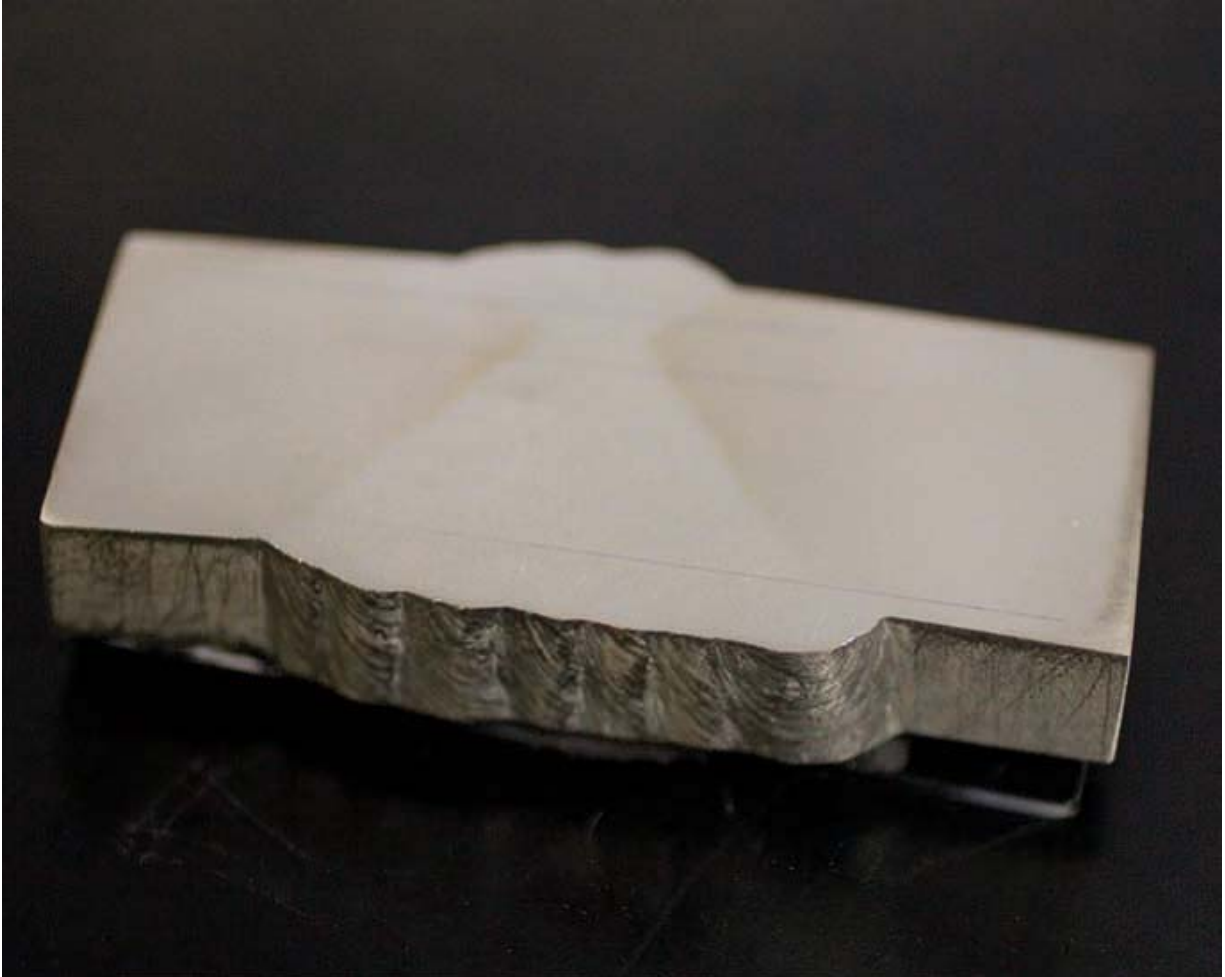


Photo 66 Macro section of multi-run butt weld with basic electrodes, 40mm plate.

Fillet welds

These are, unlike butt welds, of the 'non-integral' type, and therefore require sizing. Fillet sizes are normally specified by their leg length, although stress calculations are based on the throat dimension. Before we can undertake any fillet weld sizing, it is necessary to understand the operation of the component and be able to estimate the failure mode. If cyclic loading (fatigue) is present then the features of the welds mentioned above, particularly stress

concentrations and residual stress, can dictate the safe application. The author has experienced failure of a replacement vessel, shortly after introduction in industrial plant, despite proof testing at four times the operational pressure. Analysis of the failure surface showed clear evidence of fatigue. As cyclic loading was outside the scope of the original specification and design, it was eventually found to have been caused by malfunctions in other parts of the plant, introducing pressure pulsations to a vessel designed purely for steady conditions. Successful service was achieved by the final replacement, of identical design, after the malfunctioning plant was identified and removed.

Table 22 Structural steel properties, BS 4360:1986.

Grade	Allowable stress (N/mm ²)	UTS (N/mm ²)	0.2% PS (N/mm ²)	Elongation (%)
43	115	430	275	22
50	160	500	355	20
55	195	550	450	19

Assuming that nothing nasty is going to creep up unexpectedly (which is the basis for most design otherwise nothing would ever get produced), then we can apply stress values from [Table 22](#) as a guide, using previous BS4360 structural steel grades, with Gr. 43 being a mild steel with ultimate stress of 430N/mm², Gr. 50 yield 500N/mm², and Gr. 55 yield 550N/mm², a carefully processed low carbon steel with superior strength and good ductility, used extensively in the welded fabrication industry.



Photo 67 Selection of single-sided fillet weld sections, macro preparation.

The generally agreed approach is to take the quoted figures as the controlling shear stress and apply them to the fillet weld shear area, which is the weld 'throat' times weld length, as in [Figure 31](#). A useful rule-of-thumb for fillet weld strength is that a $\frac{1}{4}$ in (6mm) leg length fillet will support approx 1 ton.f (1 tonne) per inch (25 mm) of weld length. Naturally, this is subject to all the limitations noted above, but it does provide a grasp of the immense potential strength of even small fillet welds, which is why they are so used and abused.

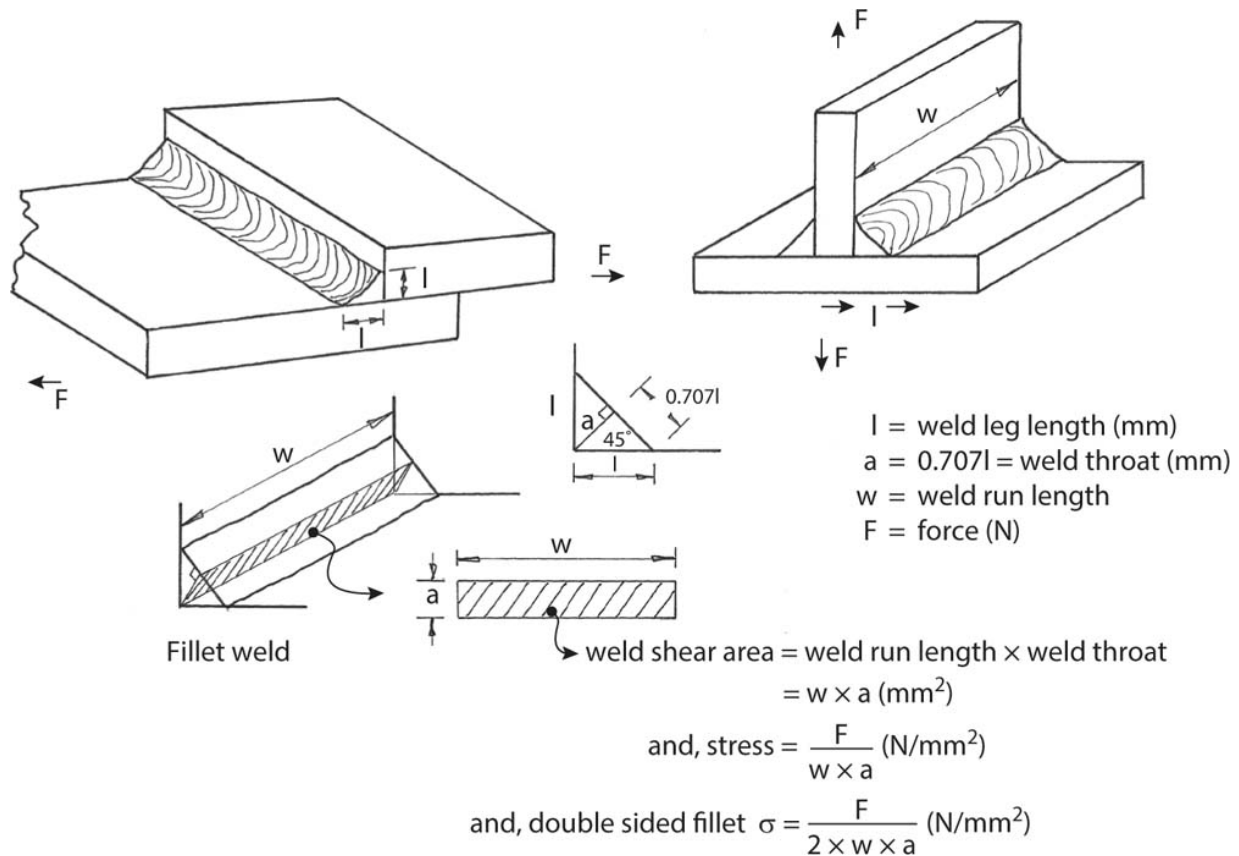


Figure 31 Stress on weldments.

A commonly used method is to size fillet welds in line with the material joined, i.e fillet leg-length equals material thickness, $l = t$. This is not recommended without an understanding of the loads being supported, and simply increasing fillet weld size can lead to problems with residual stress. Whenever possible, fillet welds should be balanced on both sides of an attachment.

There is a fundamental weakness of fillet welds where fatigue conditions occur. The unwelded material that inevitably exists between two fillets provides a prime site for crack initiation. This is the reason why full-fusion joints are often preferred, or mandatory, in pressure vessel design. For highly stressed cylinders, any brackets welded to the outside should be arranged with the weld *parallel* to the axis, as welding around the cylinder (normal to the axis), will create a stress concentration at the weld toe. Combined with the basic hoop stress being double the lengthwise stress, this provides a prime site for fatigue crack initiation.

Fillet welds in torsion can be treated in a similar manner. Note that attachments can often generate loading in combinations of torsion and bending and analysis may be difficult, in which case some experimentation may be required to test out a safe design.

Finally, care must be observed when conditions of high restraint occur, such as heavy wall sections or opposing welds. There are many examples of failure by the insidious mechanism of hydrogen cracking, as known to the author in a large offshore crane slewing-ring, fixed to the cabin base by a very large weldment. Extensive non-destructive testing showed no problems immediately after welding and the job was duly completed, but failed on load testing as the trapped hydrogen had slowly evolved from small internal cracks to unrepairable surface-breaking defects, resulting in the whole construction being consigned for scrap, at great cost. A properly designed welding procedure and fabrication control enabled safe subsequent manufacture. Fortunately, the home workshop is unlikely to encounter such large-scale effects but it is quite possible to find restraint from opposing welds, such as a fabricated hand wheel, where completion of spoke welding is rendered unworkable.

4 Material selection and limitations

MECHANICAL PROPERTIES

As designs often arise from a pre-ordained choice of materials, an existing example, or even simply the use of materials to hand, the various metals were examined individually in [Chapter 2](#). And it is often the case that a project, such as a workbench or jig, is required where its operation involves no more consideration of materials beyond their availability. A search through the stock of material will, hopefully, provide sufficient to at least begin construction.

However, for more serious tasks, or where no material is already to hand, then it is worth considering material selection from basics, taking the opportunity to review materials with respect to properties of strength, stiffness, ductility, and toughness; against the common modes of failure, deflection, ductile and brittle fracture, fatigue, and wear. These features may be critical to designs where we need to optimize weight against strength or stiffness. Some examples of these choices are outlined for a beam where suitable choice for arduous duty may involve special alloys where the ever-present cost factor may be significant, and may even be sufficient to cause a major change of plan, before the wasted expense of chasing an impossible dream.

Strength

From [Chapter 3](#), we have sufficient tools with which to characterize strength, primarily through the stress–strain curve, where we are interested in usable maximum values of stress, normally taken as the yield stress (σ_y).

Engineering materials can be categorized as:

- ◆ Minimal (low strength)
- ◆ Moderate (medium strength)
- ◆ Maximum (high strength)
- ◆ Extreme (ultra-high strength)

Minimal strength ($\sigma_y < 200 \text{ N/mm}^2$, 30 ksi)

This includes the pure elements with a face-centred cubic (fcc) structure. In unalloyed form, fcc aluminium has a very low yield strength. At $\sim 12 \text{ N/mm}^2$, for 0.1 per cent impurity content, its strength increases rapidly with greater impurity levels, to $\sim 50 \text{ N/mm}^2$. Significant alloying is required to achieve sufficient strength before serious engineering use is possible, but these materials still remain within the minimal strength range unless special ageing and/or cold working is involved (see [Chapter 5](#)).

Light alloys of the hexagonal close packed (hcp) structure, such as the magnesiums, can occupy the upper range of this category, but in sheet form they suffer from a wide range in properties when loaded in different directions, a difficulty noted in [Chapter 2](#). They can also present difficulties in fabrication by welding. The titaniums, in their essentially non-alloyed form, are characterized by the effects of small amounts of oxygen where strengths soon progress them into the 'moderate' category. However, the light alloys still benefit from a high strength-to-weight ratio, derived from their low density, despite relatively low strength.

The most important structural material in this category is mild steel, with a yield around the top of the range, and it can be work-hardened for increased values. Its properties of high ductility and low cost make it the first choice in most cases, unless corrosion resistance is important. The development of many finishes can now provide satisfactory corrosion performance, with protection by zinc plating (galvanizing) being the prime example. Where this is insufficient, then some of the aluminium alloys, such as the non-heat

treatable Al-1.5 per cent Mg, strain hardened, may compete with mild steel. The heat treatable aluminium alloys generally range between the top of the 'minimal' and bottom of the 'moderate' categories.

The Cu-base alloys, apart from the aluminium bronzes and some brasses, are generally confined to this range, and rely on their corrosion resistance and good electrical conductivity for applications, despite higher cost. Machinability is also high, but that is not uncommon in most of the materials in this category.

Similarly, apart from cast iron and some cast aluminiums, most of the metals in this range exhibit good ductility and workability, -naturally forming most of the material employed by the home-engineer, unless service conditions are arduous or application under stress is safety-critical.

Table 23 Reynolds lightweight tubing dimensions.

Reynolds tube reference no.	953/931		631/853		525/725		TITANIUM	
Draw Bench	Butting Draw Bench	Conventional Plug Bar Draw Bench**	Butting Draw Bench	Conventional Plug Bar Draw Bench	Butting Draw Bench	Conventional Plug Bar Draw Bench	Butting Draw Bench	Conventional Plug Bar Draw Bench
Maximum length (mm)	740	*	740	3000*	740	Large range* (up to 4000)	740	Various sizes* (up to 3000)
OD range (mm)	16–38.1	16–38.1	22.2–44.5	19.0–44.5*	19–41.3	Various sizes* (16–44.5)	16–41.3	Various sizes* (16–44.5)
Wall thickness range (mm)	0.7–0.4	0.7*	0.4–2.2	0.9–2.5*	0.6–1.2	Various sizes* (0.7–0.9)	0.65–1.00	Various sizes* (0.7–0.9)

* Only used for custom applications and subject to Annealing/Heat-Treatment restrictions

** Some 1.00 mm wall thickness may be available

Moderate strength ($\sigma_y \sim 200\text{--}700 \text{ N/mm}^2$, 30–100 ksi)

Of the light metals, only commercial purity titanium (Ti-O) reaches this level, at the lower end. The α - and β -phase Ti alloys are capable of competing well up this range, but their high cost and special heat

treatments usually confine their use to critical aerospace and sports applications or where cost limitations are less sensitive.

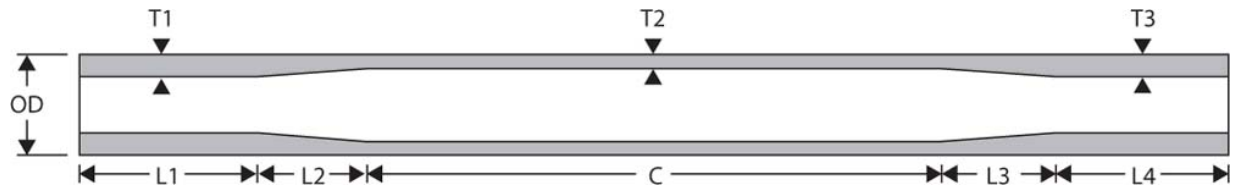
The 2000 and 7000 series heat treatable aluminium alloys have strength capabilities, in the fully heat treated condition, into the middle of this category. Combined with their relatively low cost and low density they are extensively used for weight-sensitive structures like aircraft. They are, however, not weldable and this confines their construction to mechanical joining, by riveting or adhesive bonding.

There are some Cu-base alloys, such as Cu-Ni-Si or Cu-Sn springs, which can be strain-hardened and thus fall into the middle of this range, but these are the exceptions.

The most significant alloys in this category are the steels. These include the 'high strength low alloy' (HSLA) steels, based on mild steel. With minimal alloying additions and low carbon, they are also known as 'micro-alloyed steels'. Careful compositional control, rolling and quenching treatments can produce strengths well up this strength range, at relatively low cost, making them attractive for volume car production and structural work at relatively high stresses. Values of yield from the bottom to top of this range can be designed into these steels and their use has extended to the petro-chemical industry and pressure vessels. However, they are far less tolerant of fusion welding and require properly established welding procedures to retain their strength when so joined. This has led to the practice of using these steels in hand-made cycle frames, joined with low temperature fillers such as silver soldering, through long-established lug designs. Similarly, butt joining of hand-made motorcycle frames has often used bronze welding in order to mitigate loss of parent metal properties. But often these have not been subjected to proper joint testing or metallurgical examination and an immaculate appearance may be confused with intrinsic strength, without proper-supporting experimental data.

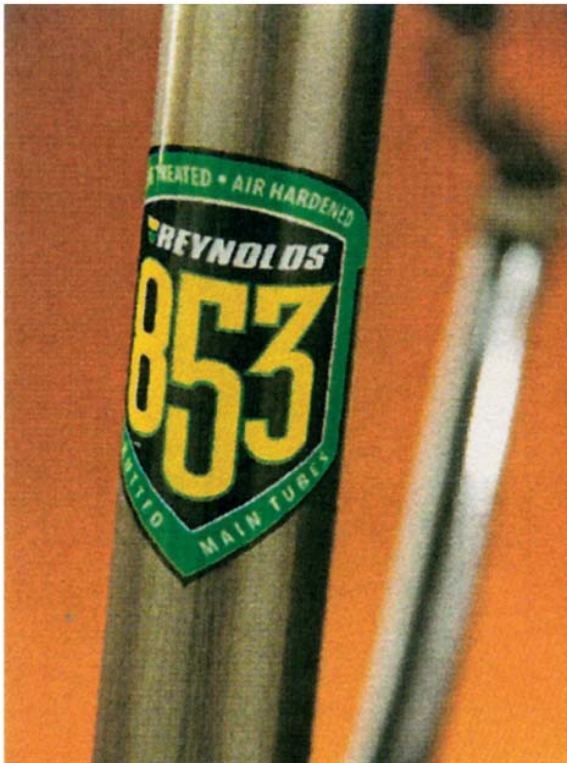
Butted Tubing (VWT - Variable Wall Thickness)

Invented and patented by Reynolds in 1898, this process can be applied to all Reynolds materials and provides a key performance – critical advantage – strength where it is needed most at the fabricated joints



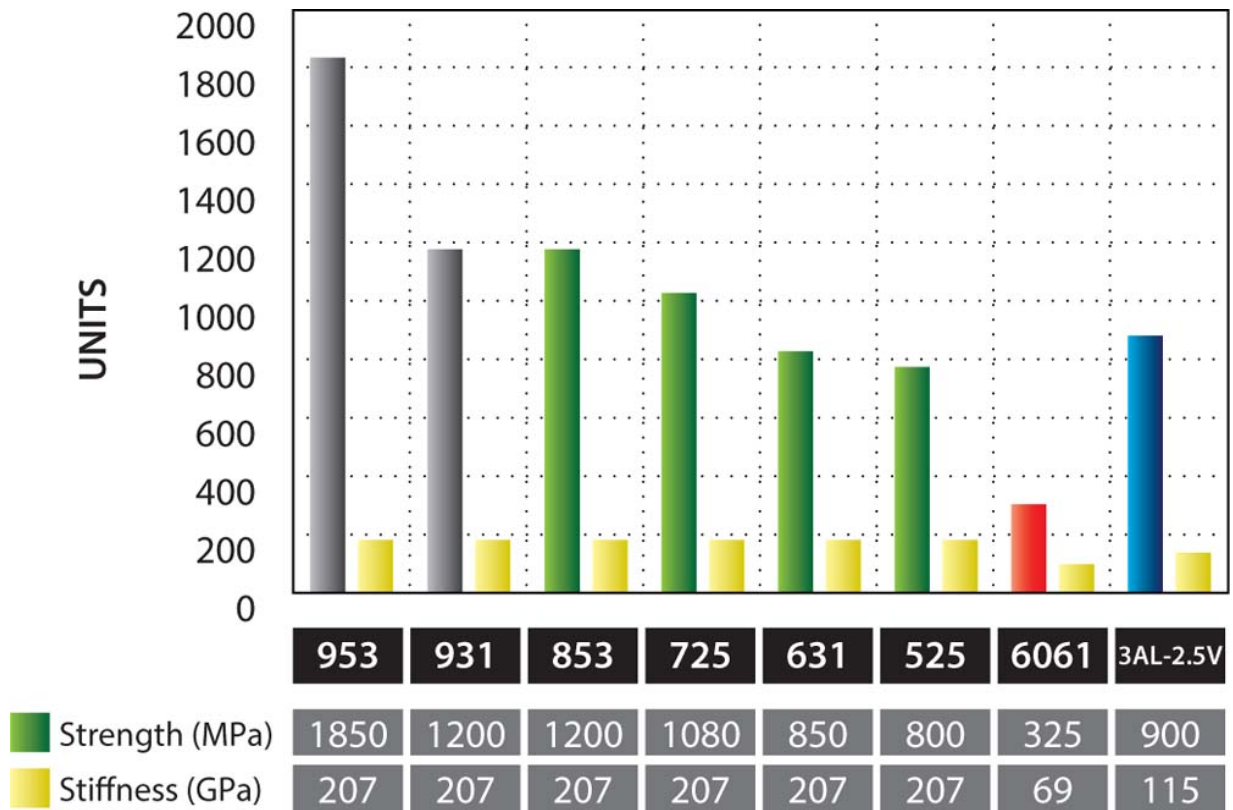
Tubes are cold drawn through a die over a shaped mandrel, so that the tube has a constant OD and variable ID

The wall thickness profile options (single, double, triple and bulge butt) depend on material, process route options and inter-pass annealing



Reynolds - materials

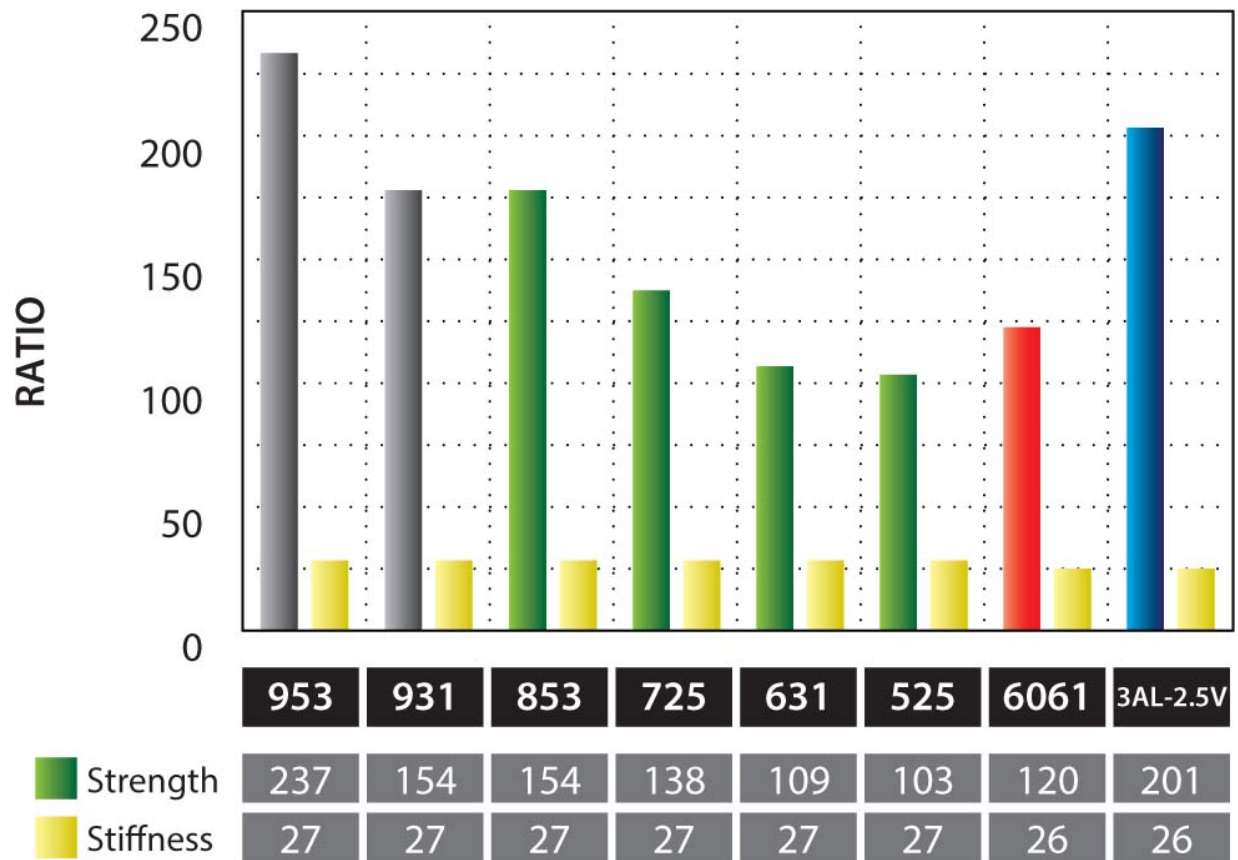
Strength and stiffness values



Typical comparative values illustrated for strength (MPa) and stiffness (GPa)

Specific material properties

Strength-to-weight and stiffness-to-weight



Figures indicated are subject to cold drawing methods and relevant heat treatment options

Typical Joint Strengths

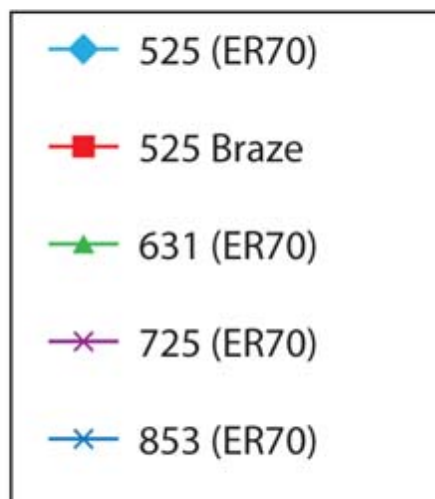
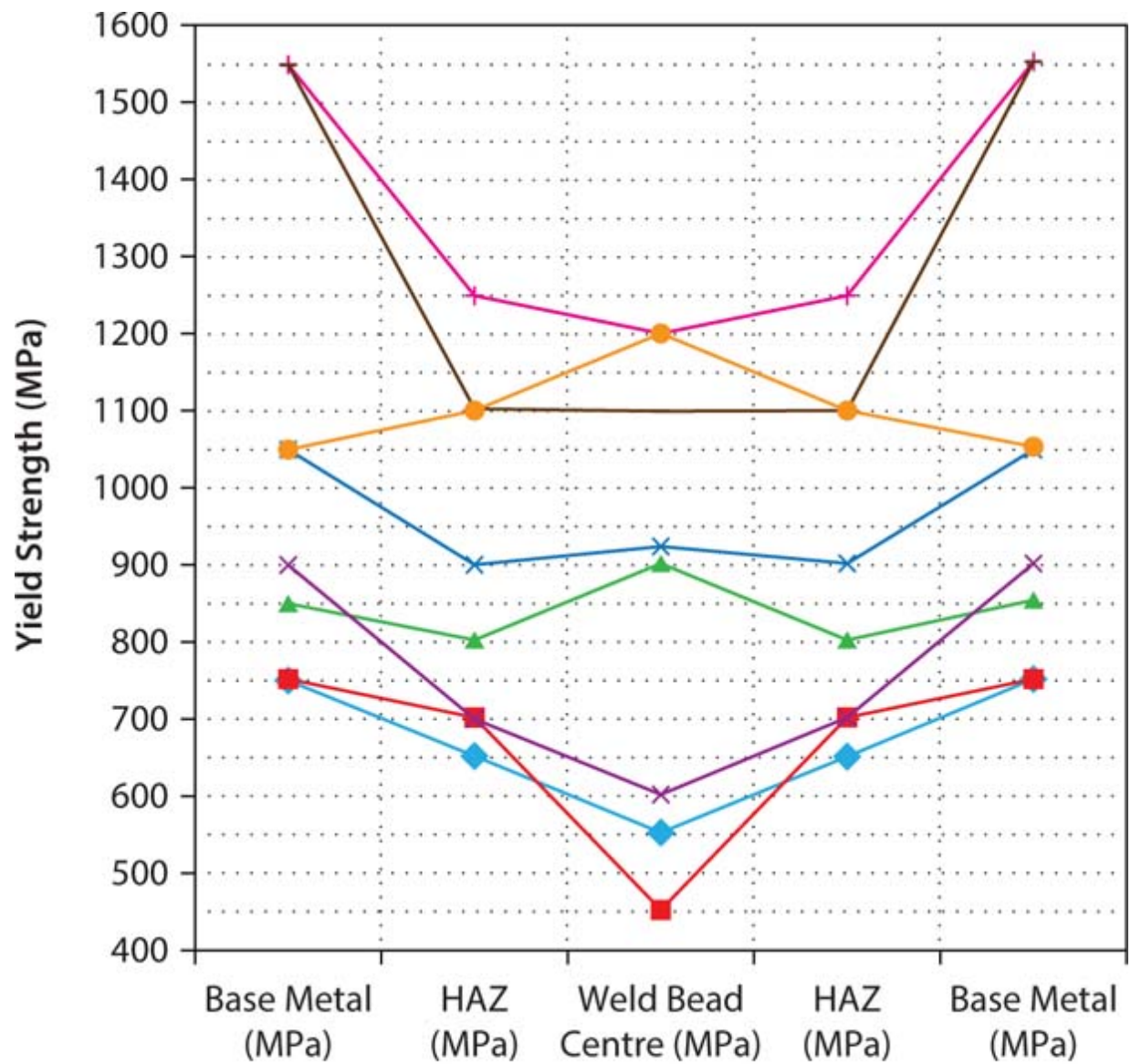




Table 24 Mechanical properties of Reynolds lightweight tubing.

Reynolds tube reference no.	853	631	725	525	953	931
Mechanical Properties						
UTS (min) (MPa)	1200	800	1050	750	1650	1100
YS (min) (MPa)	1000	650	800	600	1450	1000
E% min	8	10	8	10	8	8
Poisson's Ratio	0.34	0.34	0.34	0.34	0.30	0.32

In the UK, Reynolds Tube Ltd, established in 1898 to commercialize their patented tube-butting process, developed the famous '531' in 1935 as tube originally intended for use in competition cycle frames. By the 1960s, 531-framed bikes had achieved a remarkable run of twenty-four wins from twenty-five Tour de France races. The butting process provides minimum wall thickness with an increase near the ends, enabling joining without loss of strength in the structure.

During the past decades this firm has developed an extensive range of tubing for the specialist, high-performance market. This illustrates the remarkable properties available in thin tubular form, their strength extending from the bottom of the 'medium strength' category, upwards. Grade 725 is based on 4130 low alloy steel, with two special stainless steels that fall into the 'maximum' and 'extreme' categories.

With these developments in high integrity tubing for the specialist low volume industrial, or even up-market home workshop, it is appropriate to move beyond the well established practice of bronze

welding or silver soldering lugs for frames. In order to capitalize on these strength advances, the use of TIG (tungsten inert gas) is more appropriate, using butt joints. Whilst these types of materials have been around for some time, it is perhaps surprising that they are only recently being more fully exploited for their potential in competing with the currently fashionable fibre-reinforced materials. Makers of these high performance materials tend to be wary about advertising chemical compositions, as they are the product of significant research and development programmes, but they will assist with fabrication advice to the informed user. Even the famous 531 tubing is not generally recognized as a C-Mn-Mo alloy, which is essentially good old EN16 – one of the handful of staple favourites that make a good starting point in any general material selection from bike frames to gearbox internals.

The low-alloy steels are capable of providing strengths from the middle to top of this category. In order to achieve these values it is necessary for heat treatment, such as by the classical means of quenching and tempering, where the very hard, but low ductility martensite is produced by the rapid quench, and then tempered to improve toughness at some expense of strength. These materials are used widely in highly stressed products like vehicle power trains, and normally produced from bar-stock. Whilst the base metal is relatively inexpensive, the heat treatment can create problems in product other than that formed from relatively small bar, and the more expensive alloys are required to achieve sufficient martensitic transformation in large sections. [Chapter 5](#) deals with producing these articles with various heat treatments, suitable for the one-off-requirement typical of the home workshop.

For industrial applications requiring this category of strength in large sections, such as heavy wall pressure vessels, the route taken is normally through micro-alloyed steels, or transformation of the high temperature austenite phase to ferrite-with-carbides, produced with additions of vanadium or molybdenum. This avoids the need for martensitic transformation, by using a mechanism for pinning dislocations, as discussed in [Chapter 5](#).

The high-alloy austenitic stainless steels, in their annealed condition, produce strengths only at the bottom of this category, but can be work-hardened to well past the middle of the range, although at the cost of greatly reduced ductility. It is also possible to enhance the strength of these alloys, to mid-range, by the addition of around 0.2 per cent N (nitrogen). The martensitic stainless steels, typically 13 per cent Cr and 0.1 per cent C, can be quenched and tempered for mid-range strength, with reasonable ductility. For all these strengthened stainless steels, joining by fusion welding requires knowledge to avoid loss of the properties derived from the strain or transformation hardening.

Maximum strength (σ_y ~700–1500 N/mm², 100–200 ksi)

Apart from a sole Cu-base alloy, precipitation hardened Cu-2 per cent Be, and the $\alpha + \beta$ titanium alloys, this category is the province of the specialized steels and high-Ni aerospace alloys. The commercial purity titaniums are not capable of joining this category and the other light alloys are left far behind.

It is relatively easy to achieve these strengths with the low-alloy steels of conventional quench and temper type, but they do usually have to contain sufficient carbon (>0.4 per cent C), making welding difficult, but quite practical with knowledge and experience. Typical of these steels is EN 24/AISI 4340, widely used in demanding power train applications. For the upper reaches of this category, resort is made to the precipitation hardening alloys, with higher Ni and additions of Cr and Co, making for expensive material.

A familiar high strength stainless steel that has been used in industry for many years is the 17-4 precipitation hardening (17.4 PH) type. These can achieve strengths in the upper range through an ageing treatment and with use of very low carbon (<0.02 per cent C), enabling the retention of good ductility for such a high strength material. There is a substantial family of alloys around this composition, as they offer additional corrosion and heat resistance, below the ageing temperatures, typically ~500°C.

The most highly developed titanium alloys, like the long established Ti-6Al-4V, can be extended well past the mid point of this category and so find significant applications in the entry section of jet engines for turbine blades where the temperature is limited. They possess a very high strength-to-weight ratio and high resistance to oxidation in these very critical applications. However, their cost and lack of lower grade uses usually makes their presence rare in the home workshop.

Normally, a more practical solution is the specialized steels, which can offer surprisingly accessible specific strengths, such as the high performance tubing met in the previous category, as Reynolds 931. This is a precipitation hardening stainless steel of the 17-4PH family, weldable for cycle frames, and providing a yield of $\sim 1000 \text{ N/mm}^2$ (150 ksi).

Extreme strength ($\sigma_y > 1500 \text{ N/mm}^2$, 200 ksi)

The 300M range of 2 Ni-Cr-Mo-V, and the die material 5 Cr-Mo-V, steels are quenched and tempered to reach into this category but in practice they present severe difficulties in machining after this heat treatment. More practical is the development of the 'maraging steels', originally with high Ni (~ 18 per cent), and small additions such as Al, Nb and Ti, to provide a *ductile* martensitic structure on quenching that could be aged to produce ultra-small particle-precipitate 'clouds' that effectively resist slip in the microstructure, but without the loss of ductility expected at such high strengths, around 2000 N/mm^2 (300 ksi) for a typical composition of Fe-18Ni-10Co-4Mo-1Ti.

There is also the important range of maraging stainless steels of which the highest grade Reynolds tube, '953', provides material of ultra-high strength for practical application in the most specialized cycle and wheelchair-racing designs. This is an Fe-Cr-Ni-Mo-Ti alloy with only 0.02 per cent C, providing fully treated thin tubing, again with strengths up to 2000 N/mm^2 (300 ksi), and manageable toughness. A practical joint strength of 1300 N/mm^2 (190 ksi) is

achievable when TIG welded under close control, but may require post-weld ageing treatment for full strength to be regained. There are other maraging steels where use is made of Mn, from 9 to 15 per cent in place of Ni, to stabilize the high temperature austenitic phase and enable full martensitic transformation during processing. These are known as the maraging-TRIP (transformation-induced-plasticity) steels, which can produce similar high strengths and facility for fabrication, an essential element in realizing an exotic design.

Stiffness

The concept of stiffness was introduced in [Chapter 3](#), in the context of the stress–strain curve where we have the intrinsic material property of Young’s Modulus (E) providing $E = \sigma/\epsilon$, allowing the strain to be calculated for a given stress. In [Chapter 3](#) we were more concerned with examining the effect of material dimensions on stiffness. For instance, a built-in beam provides the relationship: $\delta = Pl^3/3EI$, where δ is the deflection for load P .

Clearly, the deflection of the beam is proportional to the load and the beam length (cubed); and inversely proportional to the Section Modulus (as the second moment of area, I) and Young’s Modulus (E).

The route to controlling deflection for a given load and beam length can therefore be through control of the ‘ I ’ or ‘ E ’ values. Material selection needs to consider both, and their interaction, in a design which is pursuing minimum weight. We are considering only the material here, for the variation in Young’s Modulus.

In metals, E is essentially a constant for a particular alloy system, as metallurgical manipulation cannot normally affect its value, beyond ± 10 per cent, and it is generally taken as a unique value. [Table 25](#) shows values for E for the common materials, noting the low E value of the light alloys, such as aluminium, magnesium and titanium, compared to the steels. The difference in E between aluminium and steel is a factor of three, so that whilst a strong Al

alloy may support the same stress as mild steel, in a beam of the same I value the deflection will be three times that of the steel beam. This is an important factor in optimizing frame designs, for instance in cycles. In order both to increase stress and restrict deflections, the light alloy will need to have an increased I value, which increases weight. This juggling act between E , I and ρ (density) needs to be played out for optimum material selection of the cycle frame.

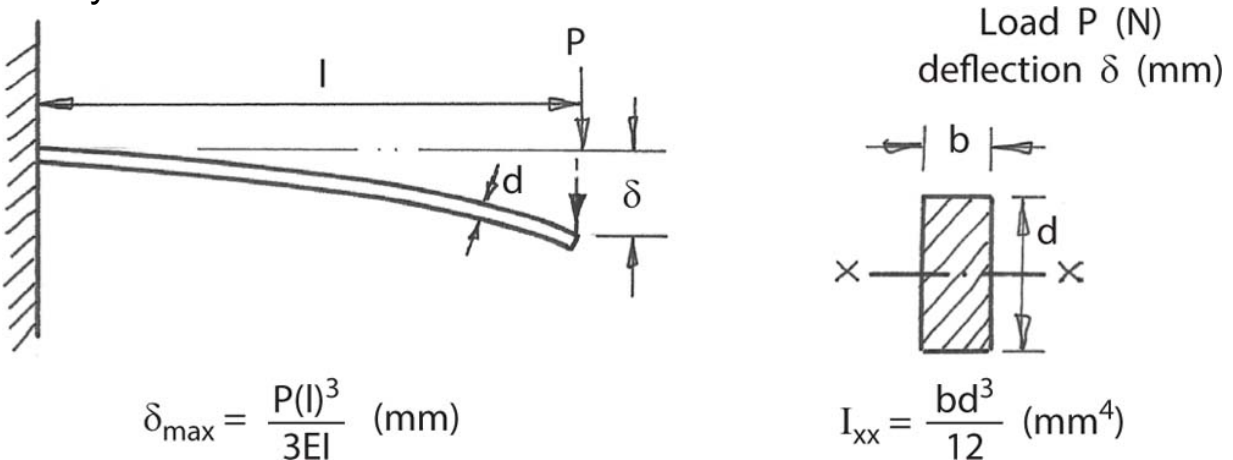


Figure 32 Beam bending (built-in beam).

Table 25 'E' – Young's modulus, approx.

Material	E (x 10 ⁶ N/mm ²)
diamond	1000
tungsten carbide	500–600
tungsten	400
chromium	290
cobalt	220
nickel	215
carbon fibre (CFRP)	70–200
iron	195

carbon and low alloy steels	210
stainless steel (austenitic)	200
cast iron	190
copper	125
titanium	116
brass and bronze	100–120
zinc	45
gold	82
aluminium	70
silver	76
tin	42
concrete	30–50
magnesium	40–45
glass fibre (GFRP)	8–46
graphite	27
wood	5–15
ice	9
nylon	2–4
polyethylene (high density)	0.6
rubber	0.05

Another feature of higher performance materials is the inevitable increase in deflections, so we are always seeking the highest E (and

l) values. As stress is proportional to strain, it is a fact often overlooked that doubling the material strength will result in doubling the deflection, if this capability is to be fully utilized. This can be an important consideration in an ultra-light frame construction, where the Reynolds tubing in the ultra-high strength 953 (a maraging stainless steel) provides a strength increase far beyond that of the original 531 (a C-Mn-Mo steel), and mild steel tubing. For each strength increase, section thickness and therefore 'I' value reduce, thereby saving weight. But care is needed to check that the strains and thus deflections are still acceptable. This marks a remarkable journey, from tubing strength levels of some 400 N/mm² (mild steel), to 800 N/mm² (531), to 1800 N/mm² (953).

The principal modes of stiffness control are: stable deflections (as considered above); energy absorption; and elastic instability. Energy absorption is affected by stiffness, as shown in the stress–strain graphs of materials with varying E values.

Whilst loads are restricted to below the yield stress, the deflections are normally small, but where they are deliberately large, as for springs, then the material choice (E) is important. Another feature of very high strength materials is the large deflections, noted above, which may have unintended consequences in 'springback', with release of the load. Where the yield is exceeded, then energy absorption becomes important, as discussed earlier, under 'Plasticity' in [Chapter 3](#).

Elastic instability is a feature common in long, slender supports, in compression. This is addressed by a number of analyses derived from the original 'Euler-strut' theory, enabling limits to be calculated to avoid the sudden buckling of a slender support. Whilst this is normally only a concern with structures like bridges, it can be encountered if long, slender connecting rods, made from the light alloys, are subjected to unusually high loads such as may occur in a model compression-ignition (diesel) engine.

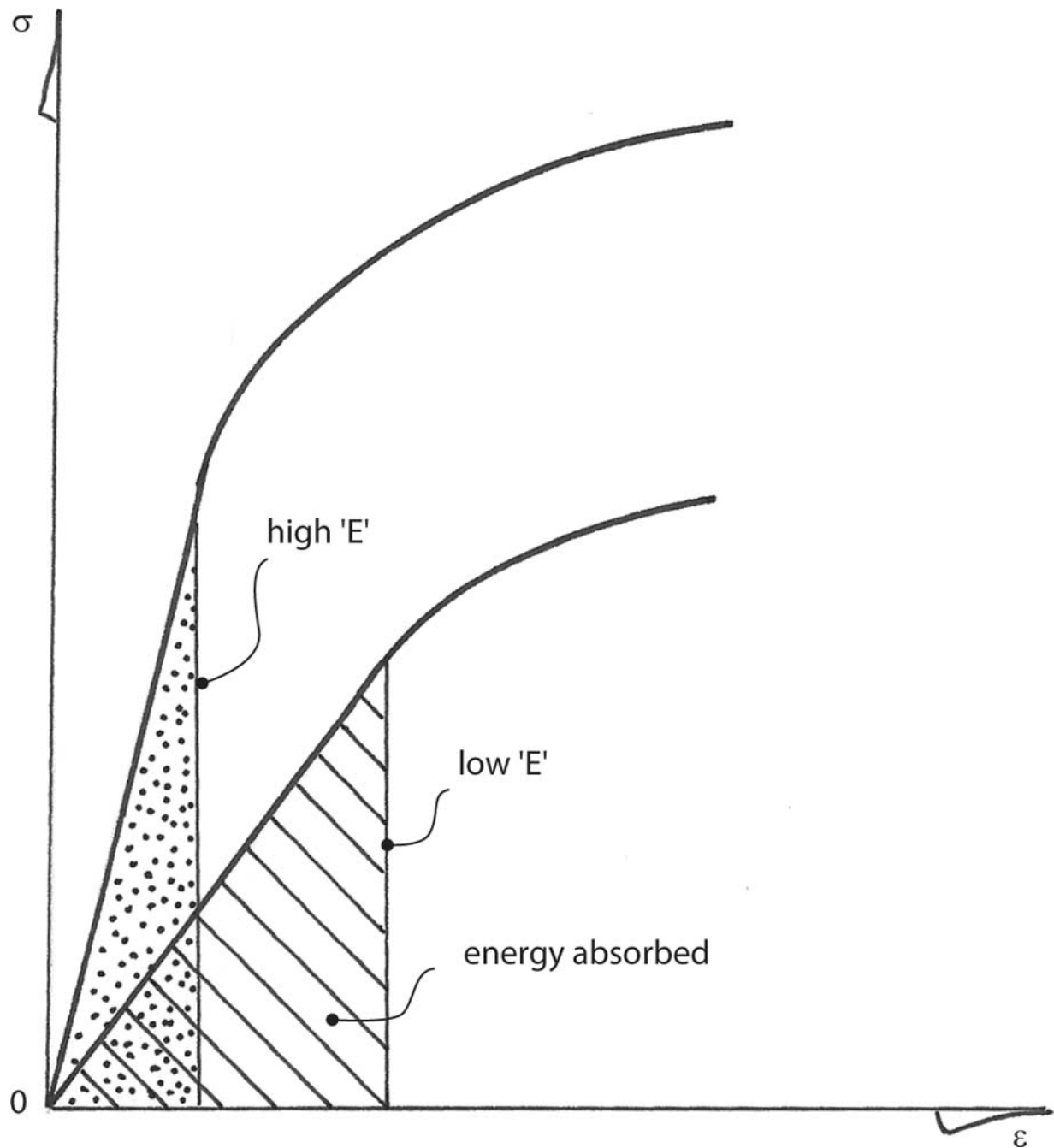


Figure 33 Energy absorption – high and low Modulus of Elasticity, E .

STRENGTH- AND STIFFNESS-BASED DESIGN

We can attempt to optimize a design limited by strength, and then stiffness, against weight. However, this has the somewhat negative virtue of illustrating the difficulty in material optimization, as it is clear that there are a large number of variations in the interaction between section modulus (I and J), and weight (volume $\times$ density), combined with variations in E and σ_y , for different material selections.

This makes it difficult to provide a single comprehensive and comprehensible formula. The following examples give a simplified analysis based on a rectangular beam cross-section, to examine the effect of choosing different material, E or σ_y values, to find the optimum design at minimum weight. However, in practice this is often hopelessly simplistic as most materials are available in various complex sections, thereby effecting vastly different I (or J) values and different weights. A fully realized optimization requires a calculation for each likely section available, typically comparing round, oval, rectangular, flange and web ('I' beam), and more; against different E values and density, for the best solution. It is, therefore, likely that experience, intuition, availability, or even chance, may provide the most likely practical means to check out a handful of the myriad of possible options, rather than trying to operate to a purely theoretical formula.

Whilst there is a certain academic interest in the rankings from these tables, it is clear that this is of limited direct application due to the limitations noted. It is notable that some materials, like the balsa, spruce and bamboo, are constrained with a simple section as a natural material, and producing a more efficient (higher I value) section is often difficult or impractical.

Therefore, the apparent attraction of the woods, like balsa, cannot normally be utilized in our engineering projects, beyond the model aircraft. These materials are also often deficient in the hardness needed to resist handling damage, and can suffer from environmental degradation, such as when contaminated with water or oil. This didn't, however, prevent their highly successful application in certain fighter aircraft in the Second World War.

The high strength and modulus of metals needs to be utilized by forming into thin and efficient sections in order to offset their higher density. The competition is with nature's remarkable composite materials, like these woods, and synthetic imitations that have now evolved in the effort to produce the best materials for weight-efficient structures, by combining low density with complex sections and lay-ups of the carbon fibre reinforced plastics (CFRPs). This is a current growth industry but a specialism outside the scope of this book, although metals can still compete successfully against the best CFRPs in projects relevant to the home engineer.

In summary, material selection needs to consider many aspects, not least cost, in order to satisfy a design for the lightest and most efficient structure. A good plan is to make a mock-up in mild steel to examine the problems with stress analysis, assembly and joining, before reaching for the exciting possibilities of the ultra high strength material, like R-953 thin wall tubing. There are plenty of material and design variables worth consideration, before costs are sunk with subsequent regret.

Table 26 Strength and elasticity vs weight.

Material	UTS (σ_y) N/mm ²	Density (ρ) g/ mm ³	E N/mm ² $\times 10^3$	σ/ρ	σ/ρ ranking	E/ ρ	E/ ρ ranking
balsa	32	0.2	1.5	160	4	8	8
bamboo	120	0.4	2.3	170	3	6	9
spruce	75	0.5	11	150	5	22	7
Al (6061)	250	2.7	72	93	8	27	1
Ti (6Al 4V)	960	4.4	116	218	2	26	5
Mg (AZ 61)	170	1.8	44	94	7	24	6
mild steel	400	7.8	210	51	9	27	1
R 531 tube	800	7.8	210	103	6	27	1
R 953 tube	1800	7.9	215	228	1	27	1

FAILURE MECHANISMS

A rational basis for design was explored in [Chapter 3](#), based on a general approach to stress analysis, where a design could be produced capable of supporting the required duty (loads) within stresses below yield and the elastic limit. However, it is traditional to apply a ‘safety factor’, colloquially known amongst engineers as a ‘factor of ignorance’, not without cause. For instance, the presence of residual stress has already been discussed at some length, potentially a major impact on the actual stress in a component often only partially, if at all, reliably quantified.

In order to assess effectively the fitness-for-purpose of a design, including its materials, it is necessary to understand modes of failure and fracture mechanisms. Experience demonstrates that most engineering failures are not due to simple yielding past the design limit, with ductile deformation before an orderly collapse as the ultimate stress is reached, following gentle plastic flow to that point. In reality, failure normally occurs, after reaching critical fatigue crack length, in a sudden and rapid manner involving ‘brittle’ fracture, at velocities up to the speed of sound in the metal. This was the scenario described earlier as the Comets exploded, with fast fracture of its thin and ductile material; and is seen when large pressure vessels fracture; or when cargo ships break up through fractures in heavy seas, with a speed that doesn’t even give the hapless crew a chance to raise a distress call. The continuing loss of

large cargo tankers on the high seas, in mysterious circumstances, is highly likely to be from this mode of failure, with the evidence residing on the ocean floor for ever.

This leads to the concept of material 'toughness', defined as an ability to sustain loading in the presence of defects, particularly cracks. The value of toughness is often as important as strength and stiffness, primarily because real structures will normally contain defects. Although, in the most favourable cases, these will be microscopic, in many instances they may already be significant, arising from welding or processing. Their stability is critical to safe engineering.

It was not until the latter half of the twentieth century that it eventually became possible to quantify toughness as an engineering property, and apply it in a scientific manner to the real world of structures and products. Before then, failures of an unexpected and often catastrophic nature were not uncommon; their avoidance was due to three major features:

1. Factors of safety were often very high, reducing operational stresses, but incurring the weight penalty of oversized parts.
2. Relatively thin and low-strength material was the norm, permitting local yielding, but precluding the safe deployment of heavy-wall construction, such as is required in modern nuclear power - stations.
3. Structures and vessels were joined by mechanical means, such as bolting or riveting. This has now largely been replaced with welding, producing monolithic designs which do not have the crack-arresting facilities offered by amalgams of individual components.

With the development of a quantitative understanding of toughness, leading to the discipline of 'fracture mechanics' where the fracture toughness of a material can be evaluated, it has become possible to reduce factors of safety, and thereby reduce weight to offer designs of an elegance previously beyond reach. This

is relevant not only in structures where weight is critical, such as aircraft or high performance vehicles, but also in the evaluation of growing cracks in offshore oil-field production platforms, critical in their safe operation where too early shutdown of operations leads to losses of millions of pounds sterling per day.

The practice of using thin material permits the best chance of it behaving in a ductile and yielding manner, due to the concept of 'plane-stress', with through-thickness stress restraint insufficient to produce the 'plane-strain' condition. In the latter case the stress at the crack tip cannot develop an envelope of yielding across the section. This plane strain condition creates the most disadvantageous state, offering the best opportunity for the crack to race across the remaining material. However, where a design dictates that a thick section is required, a lack of knowledge of fracture mechanics exposes the designer to the onset of brittle fracture should a crack of critical length exist, or develop in service.



Photo 68 Charpy impact test specimens, surface deformation (shear lips) indicates high ductility and impact values.

Finally, the increasing use of welding, replacing plates connected with bolts or rivets, removes the crack-arresting effect of the separate plates. This was most dramatically illustrated in well-known failures of the 'Liberty Ships'. These were produced at remarkable rates of productivity on the Eastern seaboard of the USA during the Second World War, launched into the chilly waters of the Atlantic, only to suffer brittle failure with a crack running right around the hull amidships, a classic fast fracture, leading to a total loss. This initiated a concerted effort to understand the brittle fracture of normally ductile structural steels. Charpy or Izod impact test pieces were used to demonstrate the sudden transition from ductile (high impact energy) to brittle (low impact energy) failure, which occurred in the range $+10^{\circ}\text{C}$ to 0°C for the steels used. The problem was

exacerbated because of the all-welded nature of the ships, and the inevitable welding and other defects were sufficient to precipitate unstable, rapid, crack growth along an uninterrupted path right around the hull, with the steel in its low toughness, crack-sensitive, condition.



Photo 69 Charpy impact test apparatus, ready for testing.



Photo 70 Sub-zero coolant bath for Charpy impact test pieces.

Whilst Charpy impact testing was a valuable tool for evaluating the ductile to brittle transition in steel used for ship plates and has since become very widely employed as one measure of toughness, it is nevertheless of almost no direct quantitative use to the designer. It *can* be related to the performance of particular materials, of particular geometry, at certain temperatures, but only in an entirely empirical manner. It merely records a relative energy value for a 10mm² material with a defined notch, and is not a reliable guide to performance of the material at different thickness, geometry, or in the presence of a sharp crack, such as in fatigue. This means the test cannot provide a quantified parameter of toughness usable in design calculations, although widely quoted and useful as a ranking tool of different materials or processing treatments. Used primarily because of its simplicity in testing and bank of accumulated experimentation, it still provides a valuable tool for quality control of the material.



Photo 71 Testpieces in coolant bath for testing at -20°C .



Photo 72 Impact test pieces, brittle fracture, low ductility and impact values, compare [Photo 71](#).

Fracture toughness

Fracture toughness is therefore an incredibly powerful tool in understanding the behaviour of real structures, and their safe operation in the presence of cracks. It enables a valuable measure of toughness directly applicable to a design, and provides a reliable estimate of the critical crack length prior to the onset of fast fracture. It is also used to enhance the understanding of crack growth rates in fatigue, capable of the prediction of part-cracked component life. The snag is that it provides a complex field of analysis requiring extensive maths for a rigorous exploration, well beyond our scope here. However, an outline of the basic concepts should be quite

sufficient for the projects relevant to our operations (see [Bibliography](#) for further exploration).

Crack propagation mechanisms

Before progressing to the basics of fracture mechanics, it is useful to know something of the micro mechanisms of crack extension, which fall into either high- or low-energy processes. The high energy process requires more energy to advance the crack and involves *ductile tearing*. It is associated with lower yield strength materials which, as we shall see shortly, allow a larger plastic zone to develop around the crack tip. The material undergoes plastic flow in this local region, often developing voids which extend and flow into 'micro-void coalescence'. Somewhat reminiscent of pulling a spoon out of treacle, it blunts the crack, absorbing the maximum of energy until the imposed load eventually overwhelms the supporting ligament. G_c is the term given to the energy function in the creation of the two new faces of the advancing crack.

For fracture we have (trust me!): $K_I = \sigma_y (a)^{1/2} = EG_c$ (see [Figure 35](#)). Fast fracture occurs when $K_I = K_{IC}$, where K_{IC} is the plane strain fracture toughness, more of which later. However, instead of being able to use a simple 10mm square Charpy test piece, for a K_{IC} test we need to produce a specimen with thickness sufficient for plane strain conditions. Effectively, this means that the higher the toughness (K_I) and lower the yield stress, the greater the specimen thickness.

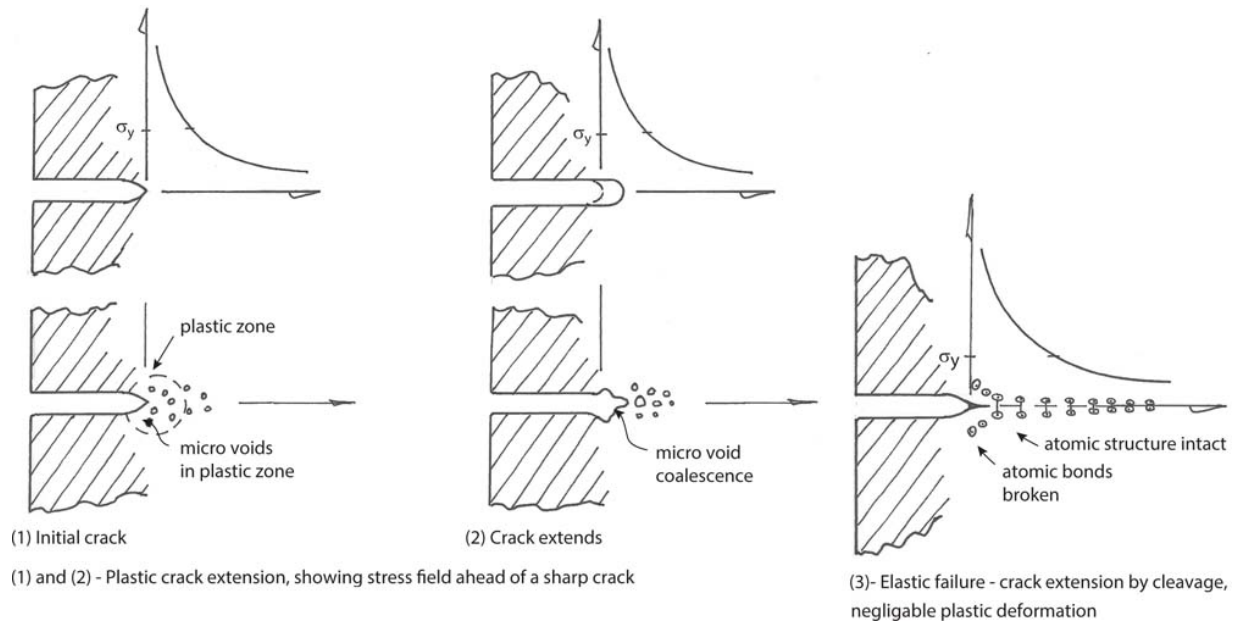


Figure 34 Sharp crack growth by ductile and non-ductile (cleavage) modes.

Table 27 Plane strain fracture toughness (K_{IC}) – specimen thickness required (t).

K_{IC} (Mnm ^{-3/2})	t (mm)
30	18
60	73
90	165
120	294

Table 28 Fracture toughness (K_I) properties.

Material	Yield strength (0.2%PS) MN/m ²	K_I MNm ^{-3/2}
mild steel	210	200
low alloy steel	500	170
maraging steel (18%)	1800	80

Ni)		
Al (2024-T4)	350	50
Al (7075-T6)	460	65
Al (7178-T6)	560	80
Ti – 6Al 4V	820	120
Ti – 4Al 4Mo 4Sn 0.5Si	1080	160
concrete		0.2 – 1.2
glass		0.2 – 0.6

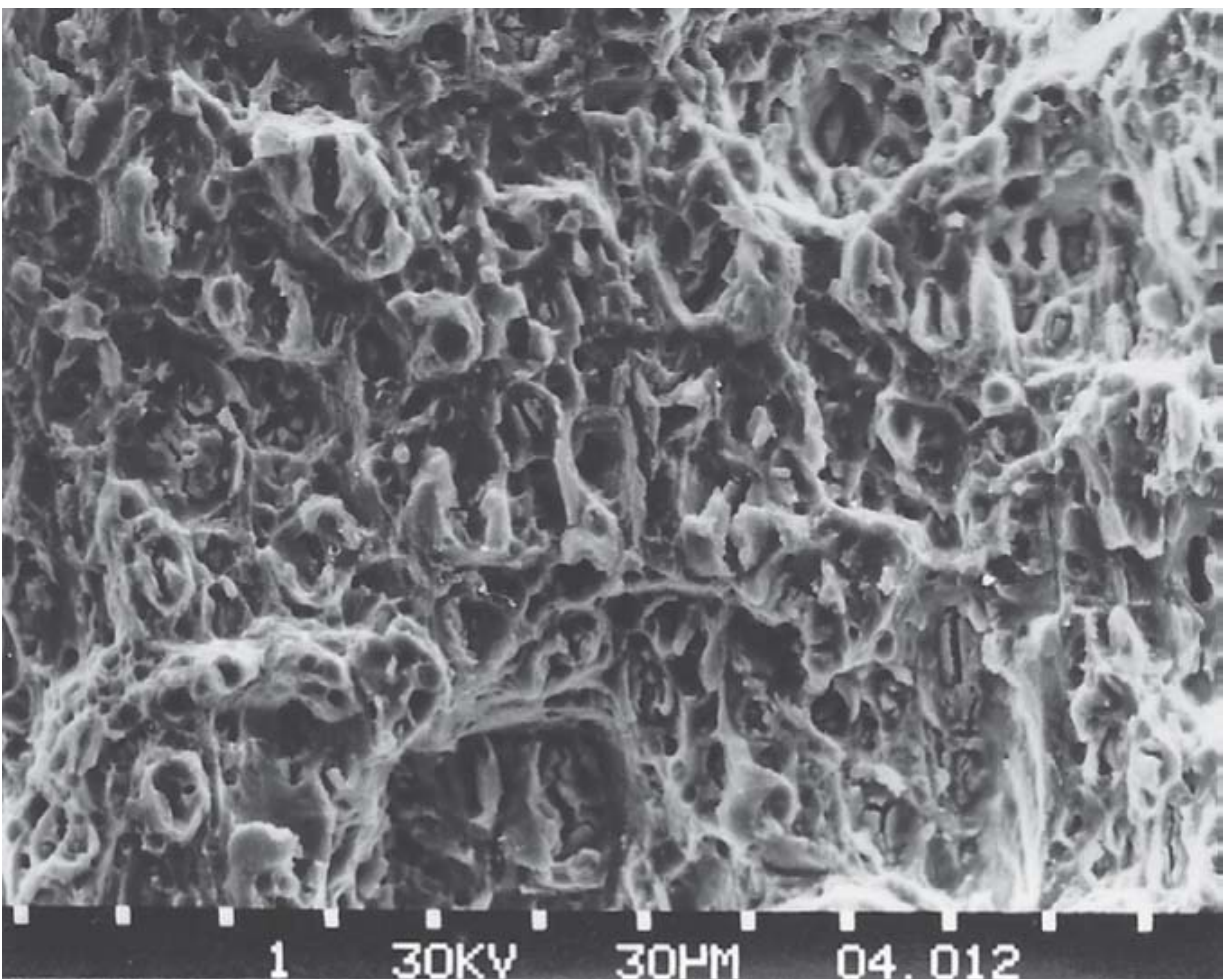
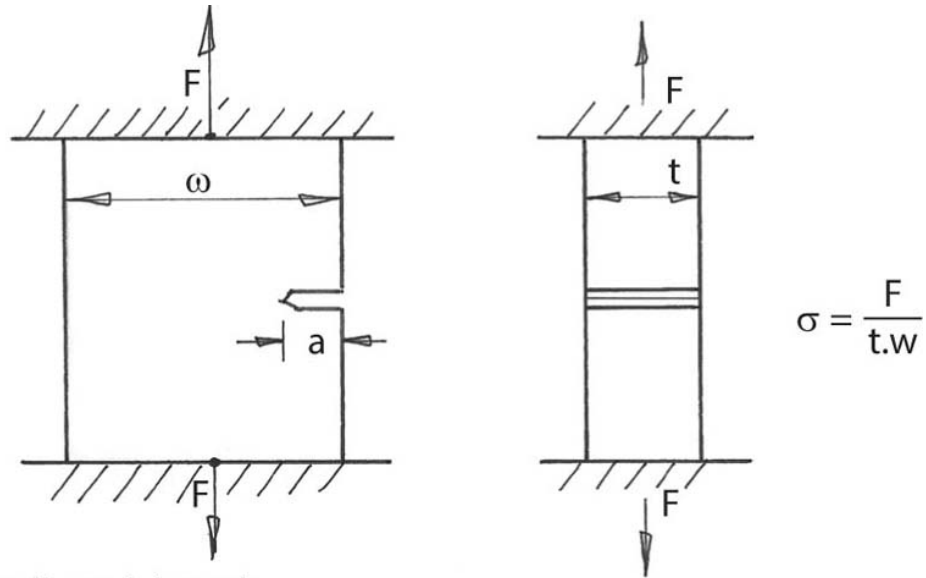


Photo 73 Ductile fracture face from SEM testpiece of 316 stainless steel, 30 microns spacing.

This leads to substantial thicknesses for the higher toughness materials, but does demonstrate how even ductile materials can behave in a brittle manner with the appropriate geometry, defect, and loading regime. The plastic zone radius at the crack tip is given by $r_Y = K_I^2 / 2(\sigma_Y)^2$, as shown in [Figures 34 and 35](#).

Ductile tearing is common in face-centred cubic (fcc) materials, which offer convenient planes for inter-atomic slip, even at lower temperatures, despite increasing strength with a reduced temperature. This is the case for fcc Cu, Al, and austenitic stainless steels, which are all devoid of a sharp ductile to brittle transition, at all practical operational temperatures. This contrasts with the body-centred cubic (bcc) and hexagonal close-packed (hcp) atomic structures. The former structure applies to mild steel, which at room temperature retains excellent atomic-level ductility, but which is lost as the temperature reduces and it encounters the ductile-to-brittle transition. Similarly for titanium alloys with an hcp structure, which have only a single plane for slip, and thus intrinsically fewer options for ductility, pre-disposing brittle failure at low temperature. (See [Chapter 5](#) for an explanation of atomic structures.)



For wide plates, small crack length,
Fracture Toughness (critical stress Intensity factor),

$$K_c = \sqrt{E \cdot G_c} \quad (G_c \text{ from standard data tables})$$

Stress Intensity Factor,

$$K_1 = \sigma \sqrt{\pi a}$$

Figure 35 Fracture mechanics testing.

Cleavage

Examination of the ductile tearing example of a fracture face (see [Photo 74](#)) shows the characteristic microscopic features of plastic flow, with rounded hollows and plucked-up peaks one might expect to see if a block of plasticine were to be pulled apart. By contrast, a brittle-fracture face presents a microscopically flat and relatively smooth surface which, resolved at the microscopic scale with the aid of the 'scanning electron microscope' (SEM), shows smooth plates of cleavage, perpendicular to the applied stress. This cleavage is evidence of the low energy process of brittle fracture, as far less energy is required to drive forward the crack, at a rate that can reach the speed of sound in the metal, for the fracture of large, highly stressed items.

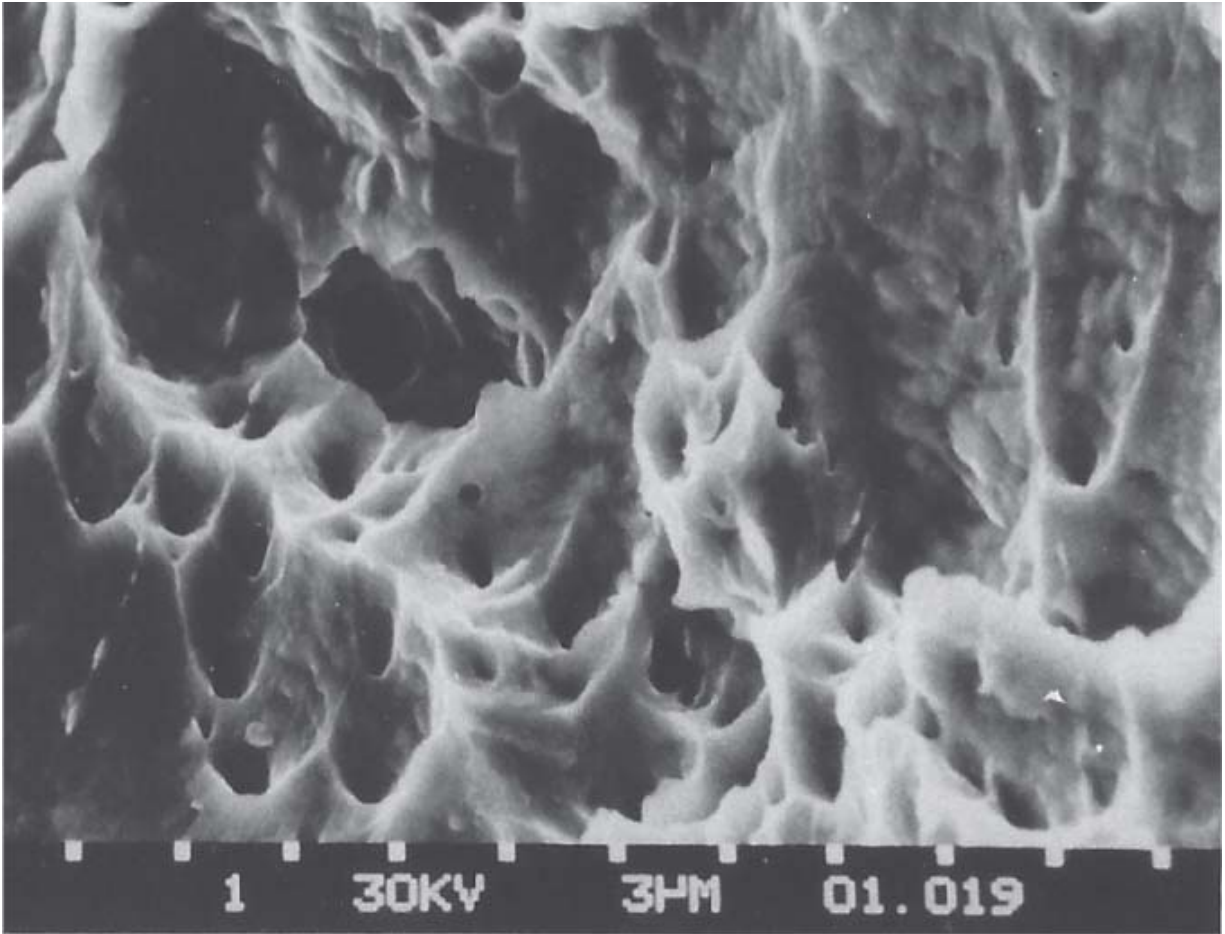


Photo 74 Ductile fracture face, SEM, 316, at 3 microns spacing.

Brittle materials and brittle fracture are familiar in everyday applications, such as non-toughened glass and ceramics. These materials are perfectly satisfactory for long-term use, provided that they are not flawed significantly or exposed to a critical stress, which is normally the case. But a sudden impact, hitting glass with a hard object or dropping a ceramic ornament, will result in fast failure by brittle fracture. The same micro-mechanism applies to steels below their ductile-to-brittle transition temperature, for the same fundamental reason of reduced ductility at the crack tip. The key to this variation in the behaviour of metals is the role of dislocations, as explored in more detail in [Chapter 5](#).

A model for the size of the plastic zone at the crack tip, which determines the mode of failure, and therefore energy absorption, is: $\sigma_{\text{local}} = \sigma + \sigma (a/2r)^{1/2}$, where σ_{local} is the local stress at the crack tip,

a is crack length, and r the radius of the plastic zone. Clearly, the smaller the plastic zone, the higher the local stress. In high yield strength and low ductility materials, this plastic zone is small enough for the local stress to exceed the inter-atomic bond strength, allowing the microstructure to literally peel apart and the crack to zip rapidly across the face, unimpeded by plastic flow and its crack-blunting effects.

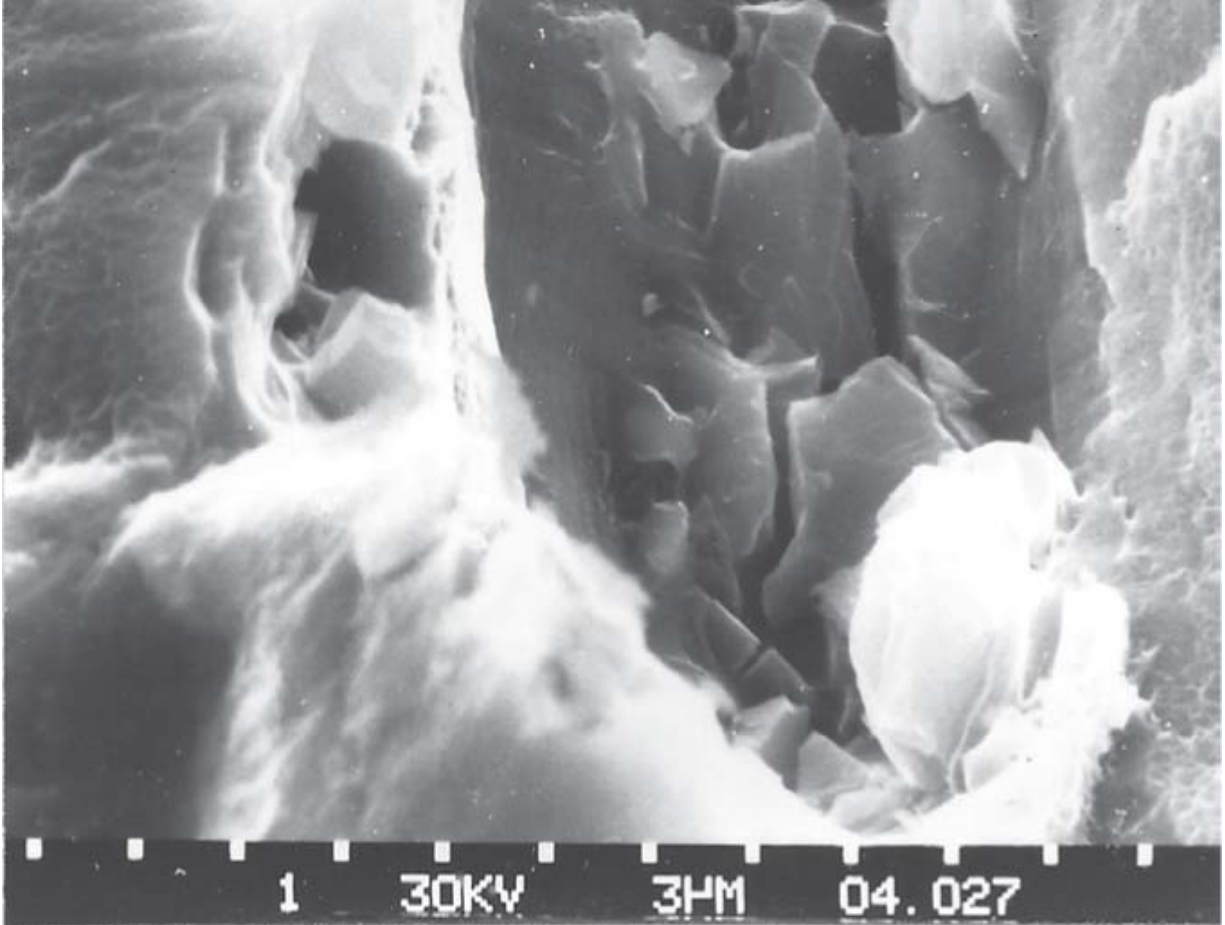


Photo 75 Cleavage fracture face details, SEM, spacing 3 microns.

Whilst it is not difficult to appreciate how the classic brittle materials, like glass and ceramics with their high yield strengths and zero plastic deformation, are subject to brittle fracture, for metals like steels at low temperature this is not so obvious. The bcc and hcp metals, ductile above room temperature, behave so because the thermal energy is sufficient for the flow of dislocations that enables the microstructure to slip and slide in its accommodation of

plastic flow, very successfully in the case of mild steel. However, as the temperature reduces, the thermal energy becomes insufficient for enough dislocation mobility, reducing crack tip ductility until the plastic zone radius shrinks below the critical size. At this point the inter-atomic bond strength, in the order of $E/15$, can be exceeded by the applied load, causing sudden fracture at substantially less than the yield stress.

The same situation arises for metals, below their ductile-to-brittle transition, as for the intrinsically brittle materials, in that failure cannot occur spontaneously. In metals, sufficient applied stress is required to drive forward the crack, in the same way that the brittle materials all around us provide perfectly satisfactory service, unless overloaded. Normal glass is usually fine for windows; mild steel is equally satisfactory if not overloaded at low temperatures, but is not the material of choice for one's ice axes on the Eiger North Face in winter!

The inability of brittle materials to provide local, crack tip, plastic flow sufficient to inhibit the onset of brittle fracture, can be improved by various schemes as demonstrated in nature. Natural materials, such as wood, are often of a fibrous composition, essentially an amalgam of strong fibres embedded in a ductile matrix. This is the principle of the new generation of high strength fibre-reinforced materials, developed from the mid-twentieth century but only in the early twenty-first century gaining significant application in the large commercial aircraft from Airbus and Boeing, or in high performance sports equipment. Thus, the high strength, low ductility, extremely small carbon fibres are embedded in a ductile, low strength, plastic matrix that serves to bond the fibres into a workable whole. This enables the very high theoretical fibre strength to be realized, permitting lay-ups that provide strength in particular directions, as required in cycle-frame bottom brackets. Such materials are undoubtedly competitors to the metals, and may offer advantages in constructing purpose-made sections with optimum strength and stiffness-to-weight ratios. However, advances in metals technology offers strong competition for the ultimate material choice: the high performance fibre reinforced materials, like CFRP (carbon fibre

reinforced plastic), present some problems with availability, cost and manufacturing methods, tending to preclude their manufacture in the home workshop, at present. The next generation of 'nano-scale' reinforcements, with the likes of graphine, whilst offering sensational performance, are still a long way from industrial application, let alone reaching our non-corporate hands.

Fracture mechanics: the basics

Fracture mechanics provides the means to understand and predict the behaviour of strong materials that contain cracks, enabling a calculation to be made of the critical length required before the onset of fast fracture. It can also be used to calculate the rate of growth of fatigue cracks in a structure, an essential element in the maintenance of all large aircraft, allowing commercial craft to keep flying by careful monitoring of slowly growing fatigue cracks, and planning for replacement in an orderly manner. Were it not for fracture mechanics, wholesale grounding of aircraft, closing of power stations, ceasing of production from offshore oil platforms, and bridge closures, would see an immediate cessation of our modern lifestyle.

FRACTURE MECHANICS OF BRITTLE MATERIALS

The ideas of Griffith were developed by G.R. Irwin of the USA, after the Second World War, to produce what we now recognize as the basis of modern fracture mechanics. This included the concept of minimum specimen thicknesses for the worst condition of plane strain, now known as the field of 'linear elastic fracture mechanics' (LEFM). The fracture toughness parameter (K_1) becomes, at the onset of fracture, designated as K_{1C} , the 'plane strain fracture toughness'. K_{1C} is a material constant, in the same way as is yield stress. K_1 , on the other hand, is the 'stress intensity factor' (usually in units of $\text{MNm}^{-3/2}$), a function of applied stress. As the applied stress rises, so does the stress intensity factor, K_1 .

Thus, we have $K_1 = \sigma_y (\pi a)^{1/2}$, as before, but this requires a factor to accommodate the effect of plate and crack sizes, Y , so that $K_1 = \sigma_y(a)^{1/2}$. Some values of Y are shown in [Figure 36](#).

A simple example serves to demonstrate an application by calculating the stress intensity factor, K_1 at the crack tip, for a 12mm crack in a large plate, with an applied

general stress of 150 MN/m^2 , or 150 N/mm^2 .

From $K_1 = \sigma_y (\pi a)^{1/2}$, where $2a = 12\text{mm}$, and $Y = 1$:

$$K_1 = 150 \times (\pi \times 6 \times 10^{-3})^{1/2}$$

$$K_1 = 20.6 \text{ MNm}^{-3/2}$$

Therefore, we have the critical value for onset of failure where the stress intensity factor equals the plane strain fracture toughness:

$$K_1 = \sigma (\pi a)^{1/2} = K_{1C}$$

An example of the crack sensitivity of a brittle material, glass, follows:

If a glass plate has tensile strength of 120 MN/m^2 ; and its strength is determined by very small surface cracks; then we can calculate their depth, a ; given (from [Figure 36](#)), $Y = 1.12$. We know, from tables of material plane-strain fracture toughness (see [Table 28](#)), that K_{1C} for glass is $\sim 0.5 \text{ MNm}^{-3/2}$.

$$\text{Thus, } K_{1C} = Y \sigma_y (\pi a)^{1/2}$$

$$0.5 = 1.12 \times 120 (\pi a)^{1/2}$$

$$a = 4 \times 10^{-6} \text{ m} = 4 \mu\text{m}$$

This shows the tiny crack required for the onset of fast fracture, and why glass cutters need only make a shallow crack to enable easy fracture in plate glass.

Returning to our earlier discussion of plastic zones around a crack tip, the expression is presented as:

$$r_y = K_1^2 / 2\pi (\sigma_y)^2$$

However, this is reliable only for the plane-stress condition, which occurs in material thin enough to avoid the lateral restraint of the thicker material where plane-strain predominates as the lowest toughness condition. For the plane-strain case we have $r_y = K_{1C}^2 / 6\pi (\sigma_y)^2$, which has been found to be satisfactory, provided $r_y < a/50$, as shown in the following example, where it is required to find the size of the plane strain plastic zone at failure and minimum crack length for LEFM to be applicable. Thus, in an aluminium alloy, where $\sigma_y = 420 \text{ MN/m}^2$ $K_{1C} = 35 \text{ MN/m}^{-3/2}$:

$$r_y = (K_{1C})^2 / 6\pi (\sigma_y)^2$$

$$r_y = (35)^2 / 6\pi (420)^2$$

$$r_y = 0.37 \text{ mm}$$

As we require $r_y > a/50$ for LEFM, we have $a > 50 r_y$, so crack length, $a > 18.5\text{mm}$ is required.

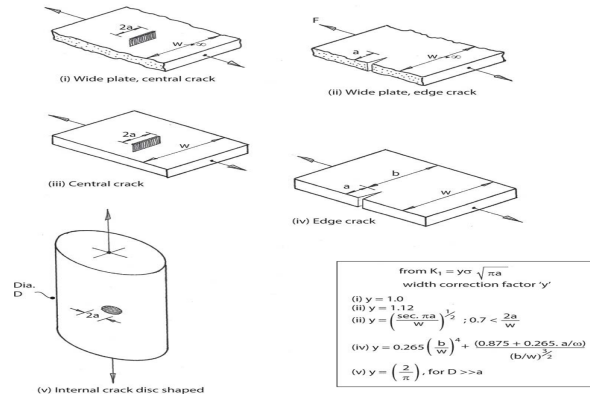


Figure 36a K_I correction factors.

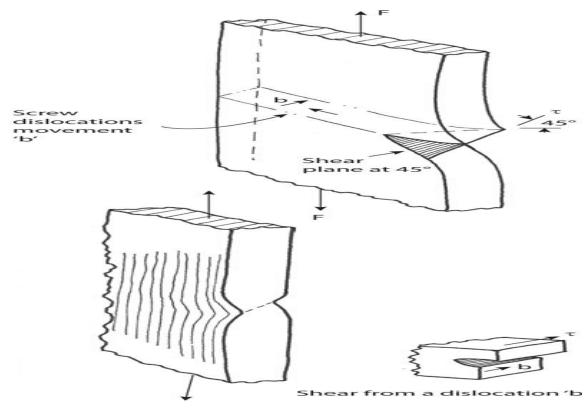


Figure 36b Thin sheet fracture (K_{III} mode).

Arguably, fracture mechanics originates from, amongst others, the work of Dr A.A. Griffith. A brilliant young scientist, he published his seminal work *The Phenomena of Rupture and Flow in Solids* in 1921, whilst working for the RAE (Royal Aircraft Establishment) at Farnborough. This work developed the idea of the energy balance involved in creating fracture faces of a brittle material in the presence of a pre-existing crack. Griffith rose to become Director of RAE in the late 1920s, and in this capacity was sent Frank Whittle's proposal for government support of the early work on the Whittle jet engine. For reasons still not entirely clear but possibly connected with his (Griffith's) alternative jet designs, the Whittle jet was rejected for sponsorship, based on Griffith's recommendations. Fortunately, Whittle was sufficiently tenacious to press on with his

shoestring research and development effort, regardless. And, although probably delayed in its adoption, Whittle's design ultimately triumphed, becoming the base upon which all subsequent aircraft jet engines have been developed, whilst Griffith later revised his earlier views and went on to do valuable work in the jet engine field.

For materials of high toughness, there is a practical problem with test specimens, as seen in [Table 27](#), particularly with mild steel needing a test piece about the width of a luxury car! This led to the development of more complex models involving plasticity, such as 'crack tip opening displacement' (CTOD), which measures the crack opening of the familiar fracture mechanics specimen, but without the need for excessive thicknesses of LEFM. Through an evaluation of the effects of extensive plasticity, these models have been successfully applied to analyse the effects of cracks in the tougher materials widely used for real life applications.

Fatigue

The phenomenon of fatigue has arisen several times in previous discussions where failures are involved, as experience shows that this is the most common form of engineering failure. Whilst we have considered the effect on materials of the existence of cracks in the discussion of fracture mechanics, it is possible to encounter failure at values substantially less than those predicted where cyclic loading operates. Similarly, even in un-cracked components, failure can occur with cyclic loading at less than the tensile strength (σ_u), and even below the yield strength (σ_y).

There are two principal fields which commonly generate such failures, that of fatigue and corrosion, with the latter being briefly discussed in the next section. This facility for precipitating failure, at such relatively low stresses, has long created problems for the designer as it can undermine all the careful calculations of stress analysis that we examined in [Chapter 3](#). It is therefore necessary for at least a general working knowledge of fatigue if one is not to succumb to previous failures, reminiscent of Karl Marx's dictum,

‘those who fail to learn from history are condemned to repeat it...’. Fatigue failures, from the trivial to the serious, are familiar in everyday life, often occurring unexpectedly in items not apparently subjected to significant stresses. Household furniture, door keys, cycle mudguard brackets, even pen clips, are all examples of apparently mysterious failures after prolonged use. They have all been subjected to a complex regime of cyclic loading, which essentially is a reversal of loading, often about a zero mean stress. In harder worked items there may be significant mean stress, or the enigmatic residual stress that assists in the onset of fatigue. Serious engineering fatigue failures can have catastrophic consequences where they come upon the users unexpectedly, as seen earlier in the Comet disasters, and the Markham Colliery failure as described at the end of this section.

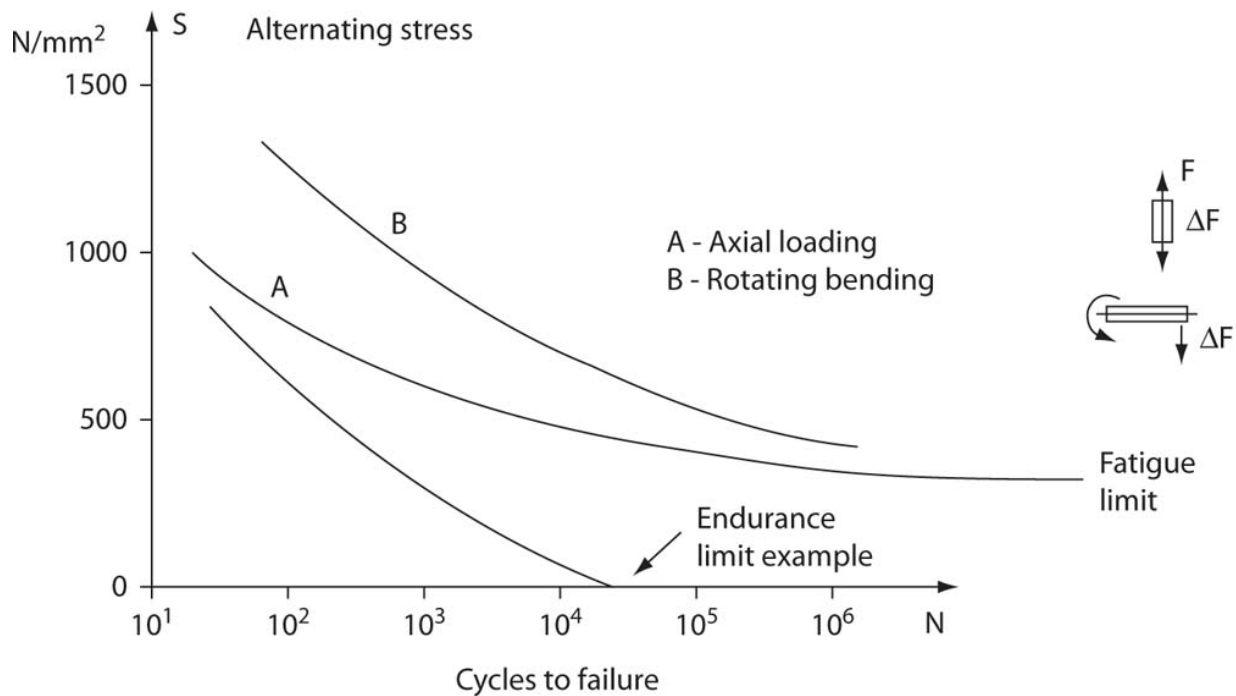


Figure 37 S–N curve (typical) for AISI 4340 steel.

This is truly an area where the holistic approach mentioned in the [Introduction](#) is important, as it identifies the potentially debilitating effect of fatigue upon an apparently sound design produced with regard to only static conditions. Microstructural and fracture

mechanics assessment is required to ensure that the design, when subjected to the varying loads virtually inevitable in real projects, is not going to fail at an otherwise unexpectedly low stress. An understanding of the loading regime, and effects like the stress concentrations that produce high local stress, is needed to predict the initiation of a new, or propagation of an existing, fatigue crack.

There are two basic fields of fatigue, as shown in [Figure 37](#): that of components without an existing crack, and those with a pre-existing crack. In the former camp we also refer to 'high cycle fatigue', conventionally set at $>10^4$ cycles to failure. For 'low cycle fatigue' we consider $<10^4$ cycles to be the appropriate life. In practice, high cycle fatigue is often associated with around $10^6 - 10^8$ cycles. This sounds a lot until we realize that for a cyclic rate, or frequency, of one cycle per second, it takes just over a day to reach 10^4 cycles, and the crankshaft of a car engine can reach this number of cycles in two minutes!

The earliest published work on fatigue is attributed to Wilhelm Albert in 1837, for investigations in the Claustal Mines. Important early work was conducted by August Whöler, who was concerned with the unexpected failure of railway axles. He set up the Whöler rotating bend test machine, to produce the familiar sinusoidal load variation, with apparatus still used today to produce the equally familiar S-N curves, as shown in [Figures 37](#) and [38](#). This is good stuff as far as it goes, but unfortunately by no means the whole story, the elucidation of which is surely worth the effort wherever we encounter cyclic loading. Whilst rotating bending is clearly an important loading regime, occurring in many important situations such as crankshafts in internal combustion engines, there are many other types, such as axial loading. These require more sophisticated equipment, near simulation, or full-scale testing as was used on the Comet failures, setting a precedent for future commercial aircraft type-proving.

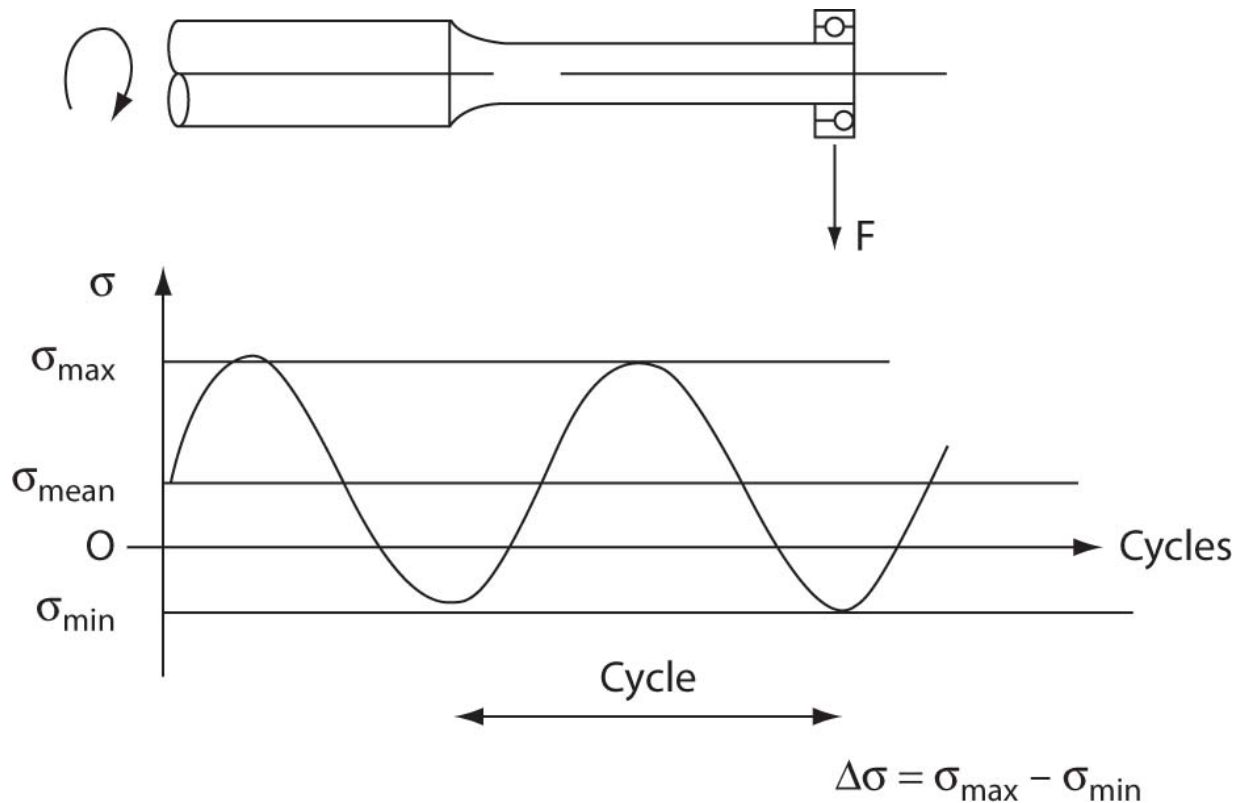


Figure 38 Fatigue: Rotating Bending.

Another source of cyclic loading, often overlooked or underestimated, is vibration. As it often occurs at high frequency and low amplitude, it is easy to underestimate the impact, such as in welded plates where high-cycle fatigue can accumulate $>10^6$ cycles in a few hours. The author has experienced such failures in simple constructions containing silencers for industrial equipment, cycling only up to 1 bar, at motor speeds of ~ 1000 rev/min. As parts of the frame resonated, all attempts to repair fatigue damage were ultimately unsuccessful, and it became clear that a complete re-design was the only way to overcome a problem that never existed until the original, smaller, machines were scaled up.

Whilst it is still possible to obtain useful data for un-cracked test pieces in rotating bending, this can now be extended economically in commercial test-houses with programmable machines that can perform axial or other loading, with complex load cycles that more realistically replicate the real-life situation. These tests still generate data that describes the material life, as in the S-N diagram. Low

carbon steels and some aluminium alloys notably produce a 'fatigue limit', below which stress the material is predicted to have an infinite life. There are, however, many metals including other aluminiums, which do not show this limit, leading to a finite life. The reasons for this behaviour remain not entirely uncontroversial, as discussed later; indeed there is still some way to go before all the finer aspects of fatigue are fully resolved, albeit of only academic interest to the home engineer.

Where a crack already exists, we now have fracture mechanics to provide reliable estimates of safe use and component life; this is very much part of everyday life in keeping the wheels of modern industrial and post-industrial society turning, and aircraft in the skies. Static loading developed for fracture mechanics, combined with finite-element stress analysis (FEA), and ever-improving crack monitoring non-destructive examination (NDE), has been combined to establish safe-life predictions in the presence of cyclic loading. Although a highly developed subject, it must be remembered that tests on laboratory specimens are only models of real life. Whilst they provide valuable material data, application to real-world components depends on the accuracy of modelling the local stress, which is heavily dependent on stress concentrations and residual stress. It may be that for safety-critical designs a simulation, or even full-scale testing, is required to validate theory by strain gauge measurement.

Fatigue testing

Fatigue testing is conducted either by 'stress cycling', where a load is applied and either life or crack-growth recorded; this aims to replicate the behaviour of, for example, crankshafts or aircraft wing structures. Alternatively, testing may use 'strain cycling', where the specimen is loaded as a consequence of amplitude, or strain, such as occurs where an electromechanical arm moves between two fixed points, or a spring is deflected between 'bump stops'.

In strain cycling, reference is made to the stress–strain curve, but with an unloading and reverse cycle. Response is a function of the material, and can result in strain-hardening or strain-softening (see [Bibliography](#)).

Prediction of performance

Design: infinite life

This is an approach derived from the S-N data, typically generated from rotating-bending testing, where a clear fatigue limit can be established. It should be noted that fatigue testing is still subject to considerable experimental variability, particularly the un-cracked type of test where, very often, the greatest proportion of the life as plotted on the S-N graph is due to crack *initiation*, and the remainder to crack *propagation* (see ‘micromechanisms of fatigue’ later in this chapter). This variability requires a statistical basis for evaluation of experimental results, as has long been understood. It was one more factor in the Comet failure saga, where too much faith was put in a single, simulated, fatigue test.

It is often useful to have a rough ‘rule-of-thumb’ measure of fatigue strength, where better data is not available. This is surprisingly often, unless one is a highly experienced specialist working in a well-researched field with substantial technical backing, unlikely in the home workshop. For the popular metals, a high-cycle fatigue limit is likely to be around 40 per cent of the tensile strength, and often below the yield strength. However, this can readily be reduced to 25 per cent of tensile strength, or even less, for a given surface finish, material condition, stress concentration and, of course, residual stress. Despite this gloomy picture, there do exist methods of successfully improving fatigue, as related later.

Although the previous discussion has concentrated on simple cycles of tensile and compressive stress, about a zero mean stress, in practice many components operate at a positive (tensile) mean stress, such as in a bicycle spoke. Failures will be from fatigue either

at the nipple (threaded-end), or the cranked hub-end. The latter is rare, as it is normally strengthened to account for the additional bending stress, a feature which we meet again as a major contributory factor in the Markham Colliery disaster.

A notable point is the case of cyclic loading leading to fatigue failure in car wheel-fixing bolts. If the nuts work just loose enough to relax their clamping force on the wheel to the hub, then cyclic loading can be 'seen' by the bolt. If the tightening load is greater than that exerted by the wheel, then the bolt will see no fluctuating load, and therefore is safe from fatigue. Note, however, that gross over-tightening can result in a single-overload failure, hence the use of torque wrenches and thread condition control.

A means of coping with mean stress is the Goodman Diagram, and its modified versions ([Figure 39](#)). This relates mean stress to fatigue limit from S-N data, given by the formula:

$$\sigma_a = \sigma_{ao} (1 + \sigma_m/\sigma_u), \text{ where } \sigma_a \text{ is stress amplitude; } \sigma_m \text{ is mean stress; } \sigma_{ao} \text{ is fatigue limit at zero } \sigma_m; \text{ and } \sigma_u \text{ is tensile strength.}$$

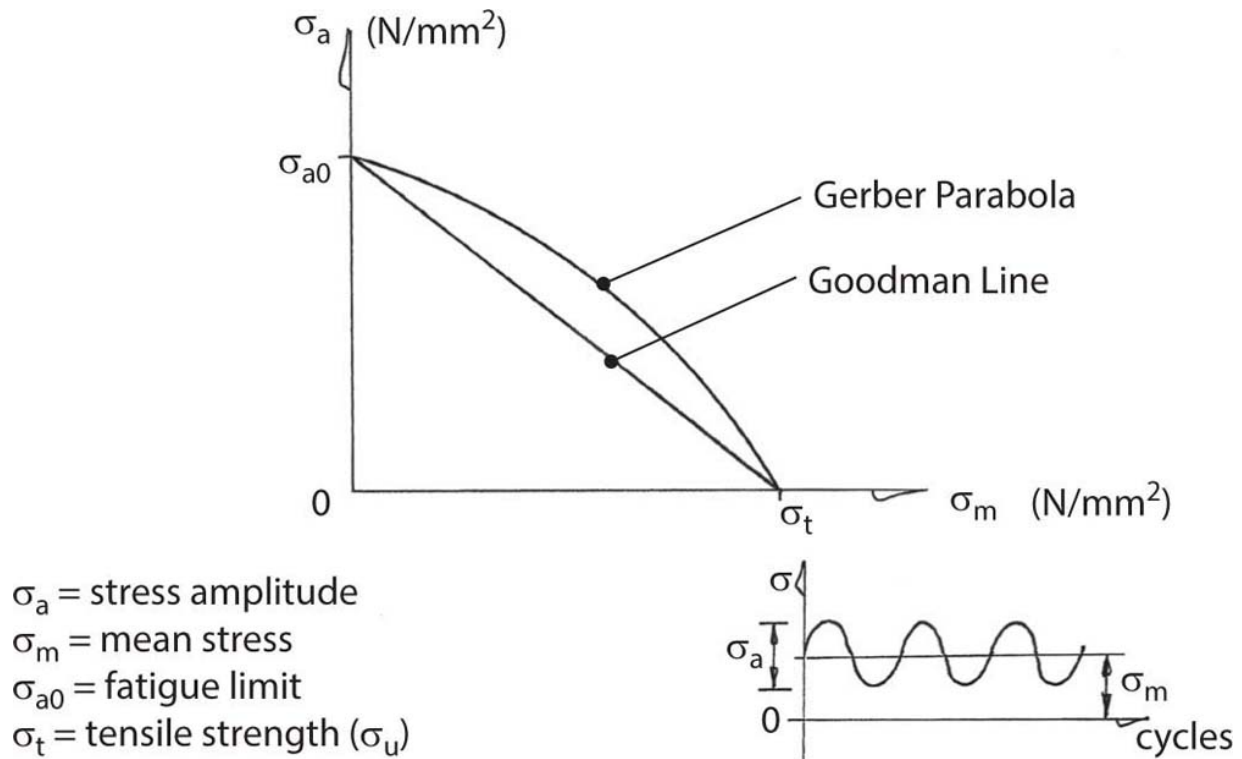
The Goodman Diagram predicts that stresses in the triangle marked will be within the fatigue limit. Practical data, however, shows safe use may be found to lie beyond the Goodman line, as modified by the Gerber Parabola. This indicates values described by: $\sigma_a = \sigma_{ao} [1 - (\sigma_m/\sigma_u)^2]$, as a better fit for [Figure 39](#).

Design: finite life

This accepts that there is no endurance limit in certain materials, which therefore must be accommodated in a safe manner, in contrast with the components designed to operate below the endurance limit. It has recently been found that even in the metals considered to have a safe stress limit for an infinite number of cycles, they may exhibit a finite life at extremely high numbers of cycles, typically $>10^{10}$ reversals; but this is not generally of practical engineering consequence.

Even since the early days of fatigue investigation, empirical rules have been applied in order to predict safe-life values for components. Various equations have been fitted to experimental data and proved very useful for guidance of likely times to failure. However, it was only with the [introduction](#) of fracture mechanics, from the mid-twentieth century, that it became possible to evaluate the effect of the critical crack length when the slow crack growth of fatigue suddenly changes to fast fracture. Furthermore, it has become possible with the fracture mechanics approach to predict fatigue crack growth rates, monitor the progress of a crack, and take appropriate action before the onset of fast fracture.

A vital aspect of fatigue life is the behaviour of the component as either un-cracked or crack-containing. In the former, most of the fatigue life is normally devoted to *initiating* a crack, after which a relatively short proportion of the life is needed to reach final overload failure. Initiation is followed by 'Stage I' crack growth, at 45° to the loading axis. Stage I follows slip planes, as it is a shearing mechanism. After ~1 or 2 mm, the crack resorts to running at right angles to the tensile load, producing the familiar flat fatigue 'Stage II' crack growth, usually referred to as 'crack propagation'.



Goodman Line:

$$\sigma_a = \sigma_{a0} \left[1 - \frac{\sigma_m}{\sigma_t} \right]$$

Gerber Parabola:

$$\sigma_a = \sigma_{a0} \left[1 - \left(\frac{\sigma_m}{\sigma_t} \right)^2 \right]$$

Figure 39 Goodman diagram for fatigue.

When examining a fatigue fracture face (see [Photos 76](#) and [77](#)), such as a rotating bending failure, there may be clear evidence of the characteristic ‘beach marks’ spreading across the flat surface until interruption by the more ragged final failure region. This is not always evident, depending on the material, geometry, and loading regime. Sometimes, as in axial fractures of stainless steel, a rather ‘chaotic’ fracture face appearance may be evident, requiring the high magnification and great depth-of-field resolution of the scanning electron microscope, for the fine detail that betrays fatigue by the presence of ‘striations’, as shown in [Photo 78](#). Whilst each striation

is caused by one cycle, not every cycle necessarily produces a striation. Nevertheless, it is theoretically possible to determine the crack growth rate by multiplying the, extremely small, striation spacing by the cycle frequency. This 'fractographic' examination can provide valuable forensic evidence to establish the difference between a failure in an aircraft due to the slow crack growth of fatigue, rather than the gross overload due to an explosion.



Photo 76 Pump shaft, 85mm dia. rotating bending fatigue failure.



Photo 77 Impeller, 115mm across blade tips, bending fatigue failure.

The importance of coping with pre-cracked structures was noted earlier, as this is the predominant case in many components. As-welded structures normally contain cracks, for instance at the toe of fillet welds, unless treated in special ways. Even apparently smooth parts are seldom entirely free from surface defects, if only microscopic scratches or machining marks, to enable them to achieve an optimum fatigue life. Similarly, stress concentrations are practically impossible to entirely avoid in practical components, where they can prove costly in limiting the fatigue life of shafts and fabrications.

Fortunately, there are a number of approaches available to minimize fatigue problems (see [Table 29](#)). In fabrications, the effect of different welded joints and weld profiles have been related to

fatigue performance. There is also the effect of residual stress in weldments which may need relieving by 'post-weld heat treatment' (PWHT). Dressing of castings and fillet weld toes is often used, and the beneficial effects of controlled shot peening, inducing surface compressive stresses, has long been recognized. If these actions are insufficient, then a material with a higher fatigue capability is required. For a smooth shaft this could be a good quality material of higher strength, but the choice requires some care, and preferably testing.

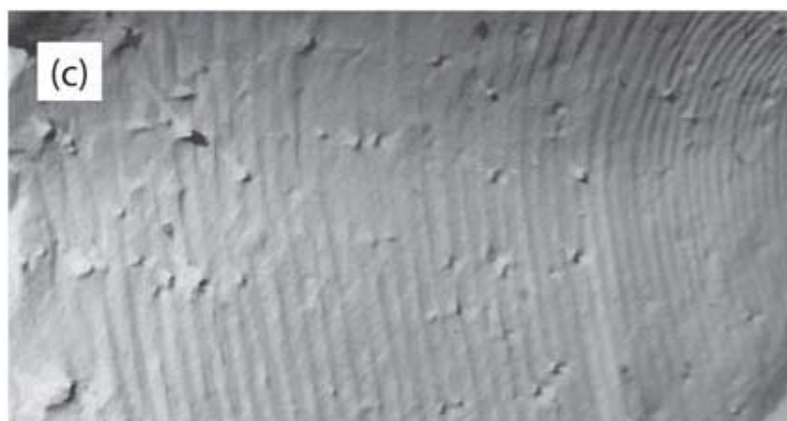
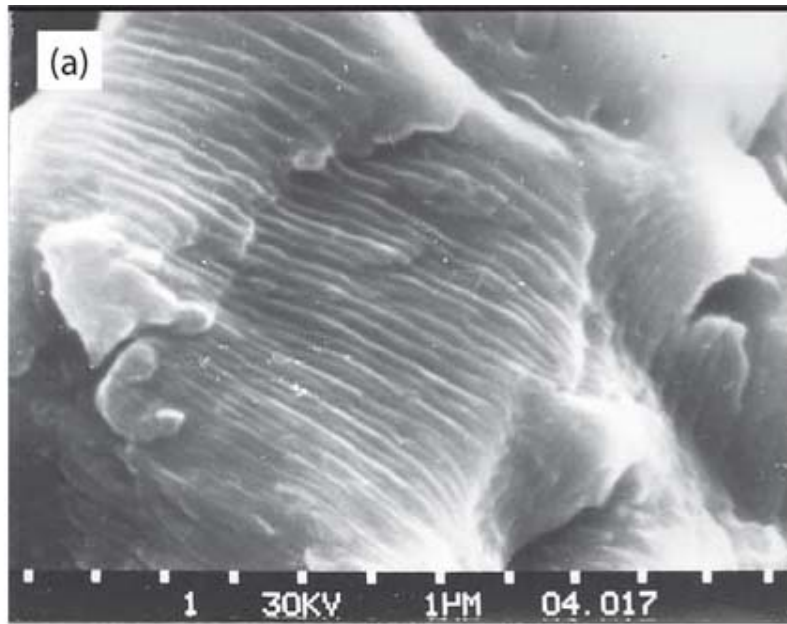




Photo 78 Fatigue fracture face details, striations clearly visible: (a) direct testpiece SEM at 1 micron spacing; (b) SEM replica, 4000x magnification; (c) SEM replica, 10000x magnification.

Table 29 Fatigue improvement from shot peening (from Shot Peening Application, Metal Improvement Co., USA, 7th Edn., 1990).

FATIGUE STRENGTH INCREASES			
Item	Before peening (ksi)	After peening (ksi)	
coil springs	75	125	
titanium joint	38	84	
gears, carburized	31	73	
weldments, HY80	26	46	
pins, 4340	71	87	

FATIGUE LIFE INCREASES			
Item	Stress (ksi)	Before peening × 10 ³ cycles to failure	After peening × 10 ³ cycles to failure
gears, carburized	80	200	30000
leaf springs	140	300	no limit
4340, polished	90	53	200
4340 nickel plated	90	39	141

Cumulative damage fatigue

A final topic in the application of fatigue damage concerns intermittent loading and realistic load spectrums. The rotating bending problem encountered with shafts operating at consistent loads, easily modelled in experiments, is often not that experienced in practice. It is a fact that fatigue damage is always accumulated, for metals do not have the self-healing capacity of living organisms and continue to degrade. Therefore, it is necessary to sum the effects of various bouts of damage, in order to predict safe life for an afflicted component.

The crack initiates and grows simply in response to the stress from the loading cycle, ignoring for the moment the effect of the environment as caused by corrosion. Despite the typically ragged appearance of the final failure surface, the surprisingly long-standing misapprehension that the metal has 'crystallized' is not accurate, as it is, of course, intrinsically a crystalline material. The only damage is that concerned with the microscopic behaviour at the crack tip, as per the next section.

A number of empirical methods have been developed to accommodate calculation of cumulative damage. The simplest and most widely used is Miner's Rule, which basically apportions each bout of damage to a proportion of the whole life, summing them to equate to unity, as:

$$n_1/N_1 + n_2/N_2 + n_3/N_3 = 1, \text{ or} \\ \sum n_i/N_i = 1$$

This can be modified by experience to include a factor-of-safety $\sum n_i/N_i = k$ (where k is a constant; n and N refer to the relevant spate of fatigue cycles; see [Bibliography](#)).

An application of Miner's Law should have figured in the fatigue predictions during the Comet disaster case, where interruption and overload cycles during testing strongly impacted on the results.

In order to apply a cumulative damage calculation, it is necessary to have loading spectrum data, and with modern electronic equipment this is increasingly accessible. For instance, data can be

logged from an accelerometer mounted in a vehicle to determine fatigue damage during a test. This data can be used with methods such as that of ‘rainflow counting’, to analyse the effect of loading frequency and stress variation, with reference to the material’s S-N curve. For Miner’s Rule damage evaluation, see [Bibliography](#).

Micromechanisms of fatigue

How, exactly, does fatigue begin and progress? Whilst much is known and may be reasonably summarized, some of the finer details and questions remain stubbornly controversial, but need not deter a sufficiently practical understanding.

Crack initiation

As mentioned previously, fatigue in uncracked metals has three distinct phases: initiation, Stage I crack growth, and Stage II crack growth. In practice, with smooth surfaces it is not usually possible to determine the difference between initiation (often known as ‘nucleation’) and Stage I cracking, as they produce microscopic features only resolvable with specialized laboratory techniques. In pre-cracked specimens, Stage I has already been superseded by the onset of Stage II cracking, so we are only concerned with the mechanisms of crack advance for every cycle of sufficient amplitude.

It is useful to the practising engineer to understand something of the, somewhat surprising, ability of smooth, perfectly finished specimens to suffer the initiation of defects that grow into Stage I, then Stage II cracks, even in strong, tough, and defect-free metals. Initiation concerns, at the most fundamental level, the atomic-level ‘shuffling’ of dislocations, the means by which metals are able to behave far beyond the simple laws of the physics of perfect crystals, where strength and stiffness are dictated from their inter-atomic bond strength, further discussed in [Chapter 5](#).

Imagine an individual atomic displacement occurs, causing a 'step' of one atomic displacement. This is known as the 'Burgers Vector'. This single dislocation movement is minute, in the order of 2 nm ($2 \times 10^{-9}\text{m}$), far beyond conventional measurement. But a single crystal can contain millions of dislocations, and similar actions can be transmitted from grain to grain across the specimen section, like a ripple down a rope, so that an approximate cumulative effect can be detected at the surface. As a very rough illustration, from the displacement of a million dislocations moving through one Burgers Vector (0.1 nm), we have a displacement of:

$$1 \times 10^6 \times 0.2 \times 10^{-9}\text{m} = 0.2\text{mm}$$

But this movement, as shear, is likely along a plane at 45° to the load axis, so the axial deflection is:

$$1/(2)^{1/2} \times 0.25\text{mm} \sim 0.15 \text{ mm or } 6 \text{ thou, which is detectable in the laboratory.}$$

Of course, this does not literally occur, but the very large number of dislocations that exist and can flow in non-brittle materials enable transfer across the section comprised of many grains and various interruptions. It is by this mechanism that localized slip, as plastic deformation, can occur below the yield stress. As shown in [Figure 40](#), this development of slip bands, with repeated cycling, can eventually produce 'persistent slip bands' which build into the extrusions and intrusions that can develop into Stage I cracks, still running along the 45° shear planes. Such behaviour has been directly observed in the laboratory, during fatigue testing of perfectly smooth specimens. With less well-finished specimens, any surface effect can provide a preferential site for this initiation phase. This may occupy very large numbers of cycles at low stress, and accommodate a large proportion of the fatigue life, producing cracks only microns deep, which are difficult to detect.

The above mechanism is only one means of nucleating cracks in smooth materials. Another common source is from the cracking of

near-surface hard inclusions, common in multi-phase, aged high-strength aluminiums, and heat-treated alloy steels. These effects operate at a very small scale, only resolvable in the electron microscope with advanced examination techniques. It arises from the local deformation of the matrix surrounding a brittle particle, causing the particle to fracture, despite the general stress being within yield.

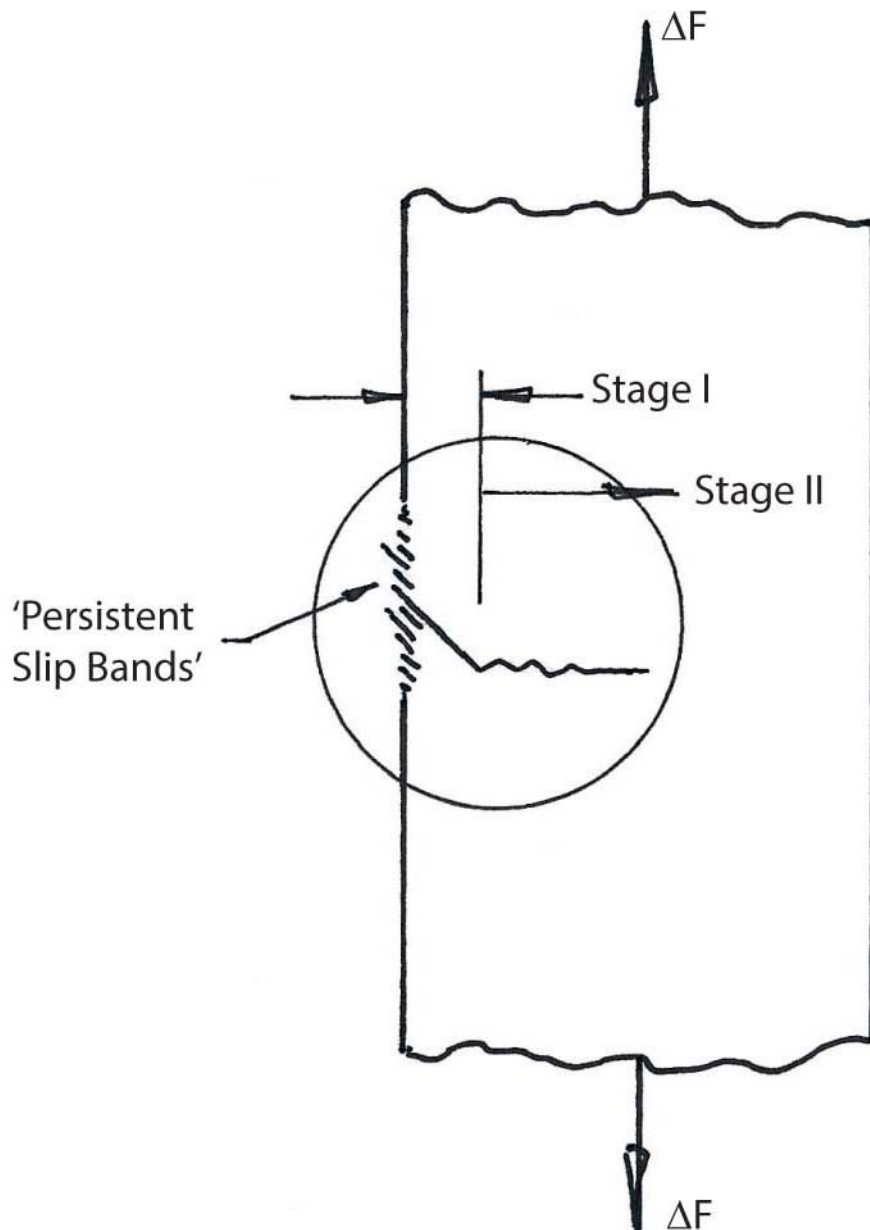
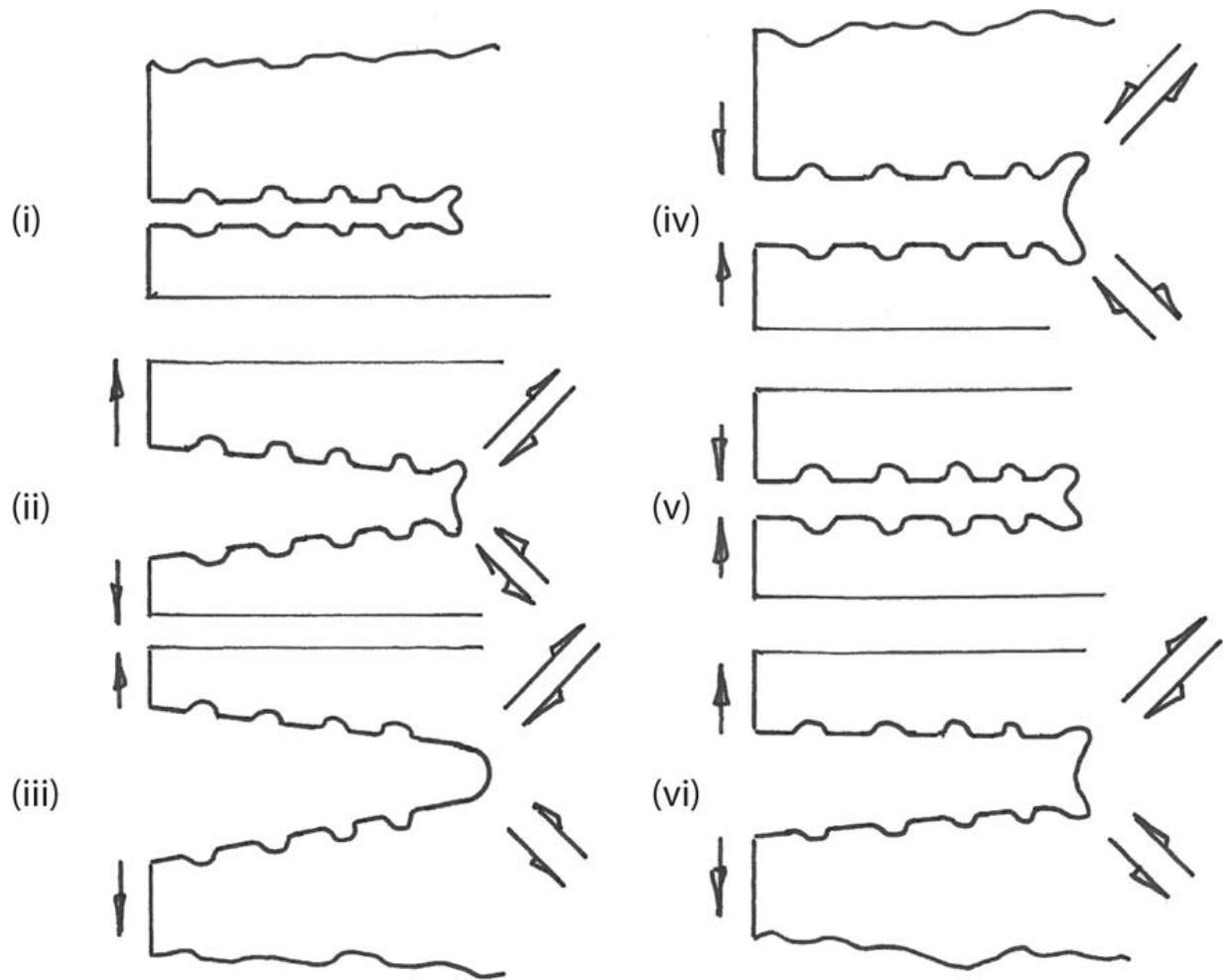


Figure 40 Fatigue crack initiation and growth.



Photo 79 Model of intrusion and extrusion from Persistent Slip Bands, for Stage I fatigue crack initiation (using profile tool).



i) To v) one cycle and striation formed
vi) new cycle begins, leading to plastic blunting and another striation.

Figure 41 Fatigue crack growth – plastic blunting and striation forming per cycle.

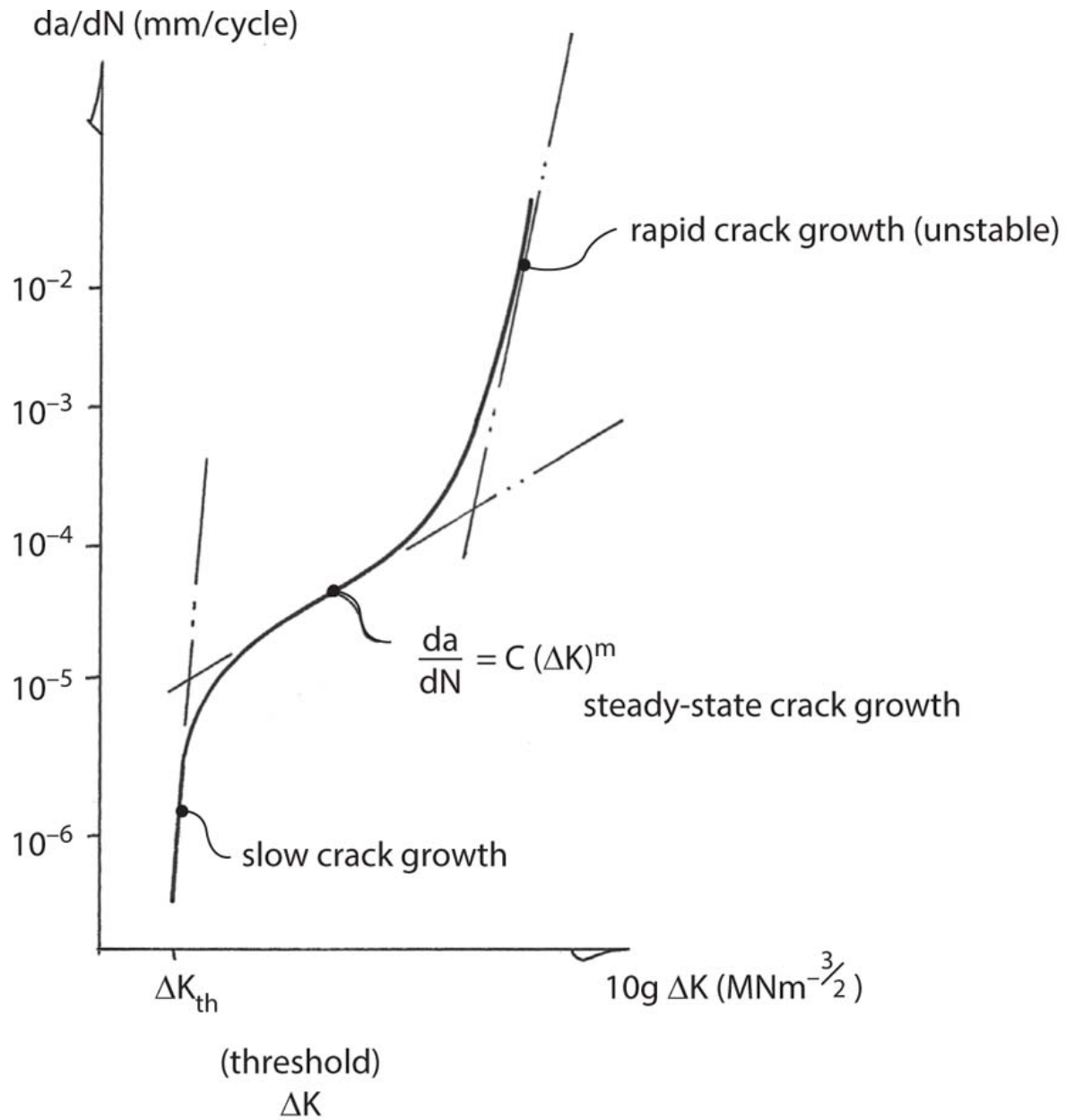
Crack propagation

After initiation and Stage I growth, the crack tends to run normal to the load axis, as a Stage II crack. For cracked specimens, this governs the entire life of the component, where the crack can advance in the sequence shown in [Figure 41](#). Provided there is sufficient stress, the crack will extend by one step every cycle. This may or may not, depending on the material, result in a striation. The more ductile materials often provide clear striations, as seen in the scanning electron microscope. In the more brittle materials this

surface feature may not be so readily observable, and evidence of fatigue more difficult to differentiate from, say, brittle fracture.

Fracture mechanics: crack growth

The application of fracture mechanics to fatigue uses the concept of variation in stress intensity factor (ΔK), in an analogous manner to the stress range of the S-N curve. But here we plot ΔK against the rate of crack growth, da/dN , which is simply the incremental crack advance per cycle. Experimental data indicates that the crack growth rate (da/dN) follows a straight-line relationship when plotted against the logarithm of ΔK , for most of the fatigue life (see [Figure 42](#)). The graph deviates at the early stage, as the threshold below which propagation does not proceed. It also increases rapidly as final failure approaches but this central, predictable, behaviour provides the key to managing crack growth in critical real life structures like commercial aircraft wing spars. Careful monitoring of the crack, combined with stress and material data to provide ΔK , is used to determine the crack growth rate and prevent the onset of fast fracture by predicting crack length before taking the unit out of service.



$\frac{da}{dN}$ (crack growth rate) vs. ΔK (stress intensity factors range)

Figure 42 Crack growth rate measurement.

Factors affecting fatigue

Having considered, in some detail, the processes involved in fatigue it is important, finally, to understand the factors which can be of use

in mitigating its effect. These fall into the categories of microstructure, surface condition, and environment.

Microstructure

A fundamental aspect of the fatigue process is that plastic flow is necessary. For initiation and Stage I damage, it may involve a microscopic process at the atomic level of dislocation interaction, assimilating persistent slip bands, leading to micro cracks. This is strongly dependent on microstructure, with high ductility, low-strength materials providing the landscape for slip bands to develop at a relatively early stage. Similarly, crack nucleation from the fracture of hard, second-phase particles is dictated by the microstructure. In this case, it is a microstructure of greater strength, as noted previously.

An improvement in yield strength, through the manipulation of strengthening mechanisms, will improve fatigue performance, but not necessarily in proportion to the strength increase. Such mechanisms include work-hardening, precipitation hardening, and solid-solution hardening. Application of chemical surface treatments is often found by experience to have a detrimental effect on fatigue. This can arise, despite the provision of an apparently smooth and flawless surface, from problems such as diffusion of hydrogen into the surface, or a complex interface with the base material, soft enough to suffer microstrains leading to local plastic flow and the emergence of persistent slip bands. Another problem can be a tensile residual stress from electroplated high strength elements, such as Cr, leading to micro-cracking.

For crack propagation, Stage II fatigue, we are concerned with a substantial crack and the resistance of its progress by plastic flow at the crack tip, absorbing energy and blunting the extreme stress concentration that promotes its advance. The microstructure required is therefore one of maximum toughness, as measured by the material fracture toughness, K_{Ic} . The microstructure needs to be strengthened by the processes noted above, whilst retaining

sufficient ductility for plastic flow at the crack tip, which deters wholesale fast fracture. Improvement of fracture toughness is therefore the key to optimizing resistance of crack propagation growth rates.

Surface condition

As fatigue cracks almost always start from a free surface, the microscopic condition is clearly of utmost importance for the smooth, un-cracked condition. Any small defect, such as machining mark, scratch, material inclusion or porosity, can provide a preferential site for crack initiation. Wherever possible, components susceptible to fatigue should have a perfectly polished surface, but this is impractical in most cases other than ultra critical, cost-insensitive components, such as racing car engine internals or suspension links.

On the macroscopic scale, stress concentration due to surface geometry is critical. This proved the origin of the Comet disaster cracks, from rivet holes, and the screw thread profile, in the Markham Colliery disaster. Any discontinuity on the surface presents a potential stress-raiser, but in practice it is often difficult or impossible to avoid. The designer can, however, be aware of the regions of greatest concern, such as the overhanging shaft which presents a maximum bending moment adjacent to a support bearing. Avoidance of stress concentrations in such a region can often be achieved by moving changes of profile or keyways to an area of lower stress. The failure by fatigue of threaded components, like the bicycle spoke or Markham Colliery trunnion, is common, and a testament to the extreme stress-raiser of the thread profile. Rolled threads, used for the former but not the latter, are always preferable as they provide a compressive residual stress and strength increase over a machined thread. Proper material selection and nut tightening can provide safe service even in critical applications, but a sound understanding is required of service conditions and fastener strengths.

One of the most important aspects of fatigue control is the use of surface treatments. The virtue of surface compressive stress has already been mentioned, and commercial processes using shot and bead blasting have proven highly successful, with spectacular improvements in fatigue life of un-cracked components. This does not, of course, influence fatigue life once a crack has progressed to Stage II growth, as for significantly pre-cracked parts. For the highly susceptible fillet-weld toes, it is common practice to carefully grind to remove the inevitable micro-cracks, or use a 'TIG re-melting' technique to minimize profile and surface defects. This can be further enhanced with localized shot-peening to induce a compressive residual stress, combating the as-welded tensile stresses.

Metallurgical surface treatments are a final method of improving fatigue life, with the role of part-transformation of low-alloy steels benefiting from the martensitic volume increase. This can create a compressive surface stress, although care must be taken to avoid 'quench-cracks' which can be disastrous even in single-load service! Hardening processes common in plain, low carbon steels include carburizing, which provides a beneficial shallow diffusion of carbon atoms into the surface. Similarly, a variety of steels heat treated for optimum toughness, may require a hard surface for wear resistance. This can be provided by a number of commercial surface treatments, such as 'nitriding', a nitrogen diffusion process. Whilst these require expensive specialist heat treatments, carburizing is an ancient process readily accessible to the home workshop, with compounds and instructions available from tool suppliers, as discussed in [Chapter 5](#).

Environment

The environment can have a significant impact on both crack nucleation and propagation. Tests have shown a pronounced effect, an improvement of around 10 per cent, simply by excluding air from the fatigue test. This is attributed to removing the moisture present

in normal air, for a medium carbon steel. During crack propagation, it has been found that the rate is particularly sensitive to the corrosive effect of various environments, and also to increased temperature where oxidation at the crack tip, combined with the reduced yield stress, can dramatically reduce fatigue life. As the environment is not normally within one's control, often the best that can be done is to be aware of the limitations, such as temperature, reducing stress, or replacement time.

Wear

Wear of engineering materials may be familiar to all, with engines and machines wearing out, and generally being a common source of failures. Nevertheless, some of the more important features of wear seem to have evaded common engineering knowledge.

Wear can be examined under the following categories:

- ◆ topography (surface profile)
- ◆ strength of material at the surface
- ◆ bond strength when surfaces meet
- ◆ deformation of the surface bond
- ◆ surface composition (oxide films and contaminants)

Surface profile and wear

If we examine surface profiles carefully, using either measuring instruments like the 'Talysurf', or prepare a cross-section for microscopic examination, even the finest finish is found to be far from perfectly smooth. The familiar finishes of fine-turning, milling, and even grinding, present on a microscopic scale an appearance akin to a mountain range. Generally, it is simply the scale of the peaks and troughs that changes from coarse to fine surface finishes. Producing a truly smooth surface is only possible by extensive polishing with a series of ever-finer diamond paste lapping compounds, and electro-polishing. Such finishes are normally

restricted to laboratory preparation of metallurgical examination samples, whilst the usual finishes available in the workshop produce deviations from perfect smoothness readily detectable without-magnification.

The business of measuring surface roughness is, at best, tricky as perceptions of fine finishes can be at variance with quantitative measurements. A fine-pitch finish can often appear smoother than one of comparable numerical value but greater pitch. Measurement of roughness needs careful control in order to get repeatable and meaningful results, and finishes are usually specified over a wide range, such as 1.6–3.2 μm CLA ('centre line average'). Consistent trace distances, and a calibration block for reference, should be employed as a minimum. Tactile comparators, produced for turned, milled and ground finishes, are normally sufficient for workshop purposes.

Where parts come together and slide, such as a shaft and bearing, the surface profile dictates how the 'peaks' will move over each other. The coarser finishes clearly offer more resistance, and greater wear rates. A geometrical parameter, relating the peak height and tip sharpness, can be used with the elastic modulus (E) and hardness (H) to determine how surfaces will traverse over each other. If the value of peak height/tip radius is low, then elastic deformation is favoured. High values of elastic modulus/hardness (E/H), implying low hardness, will favour plastic deformation. In the common ductile alloys, the peak tips will tend to deform plastically. These parameters can be used to provide a quantitative measure of materials whose surfaces will ride over one another elastically, rather than deform plastically and lock together. If the load or material strength is too high, then the peaks can fracture and be displaced in a way that can lead to locking, 'seizure', or create sites for fatigue crack initiation.

Adhesion wear

In addition to the effects of surface geometry, we need to consider the inter-atomic attraction forces. If metal surfaces are brought into intimate contact, in the absence of separation layers such as oxides or contaminants, then a metallic bond of immense strength can form, essentially no different to the strength of a grain boundary. However, in practice surface films are always present, unless actively ruptured or removed by special means. This prevents all metals joining together spontaneously, which would lead to a somewhat chaotic world! An exception, with perfect cleansing of the surfaces and exposure of the underlying microstructure, is 'explosive welding', where the enormous pressures of detonation cause the surfaces to exude a jet-stream ahead of the rapidly collapsing plates. Another example is 'friction welding' where, typically, a rotating shaft is brought into contact with a stationary part. The high friction forces disrupt the surface films, and raise the interface temperature for a strong bond, finding extensive application in high volume production, such as car driveshafts.

A considerably less dramatic effect can occur as part of a wear mechanism, where surfaces are brought together, and localized areas, 'islands', can adhere. Sufficient load, surface geometry, and disruption of surface films can then progress from the tearing-off of surface material, ultimately to the point of fusion as seizure (see [Photo 43](#)). This is an occurrence not unfamiliar to riders of racing, or even hard-pressed road, two-stroke motorcycles, when the loads between piston and cylinder overcome the tenuous lubrication, sometimes with spectacular results!

The mechanism of adhesion in wear is illustrated in [Figure 48](#), where hard and soft materials are in contact. This enables the rupture of the peaks to assist in continuing operation as the materials 'bed-in', the principle of running-in bearings that can then operate at extremely low wear rates. This leads to the well-known principle of using a hard driven material supported in a soft bearing material.

The trunnion and bearing in the Markham Colliery failure were both mild steel, an example of poor choice for operation and wear. A better option would have been a hard shaft, even case-hardened

mild steel, running in a bronze bearing, possibly of the 'Oilite' lubricant-impregnated type, with proper provision for sustained lubrication. This would have eliminated the adhesion wear evident on the trunnion shafts, allowing the low friction bearing faces to move properly, thereby eliminating the bending moment in the brake rod, a major contributing factor in the fatigue crack that led directly to the failure.

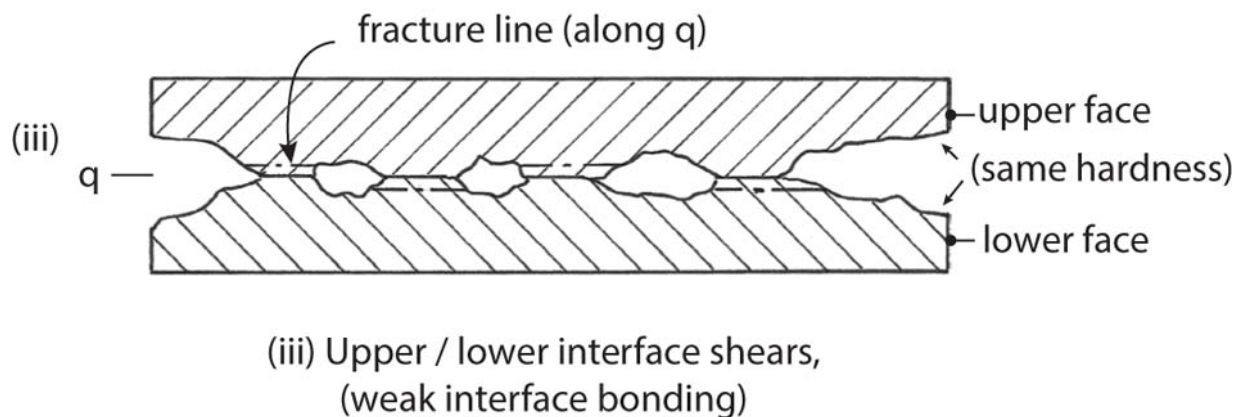
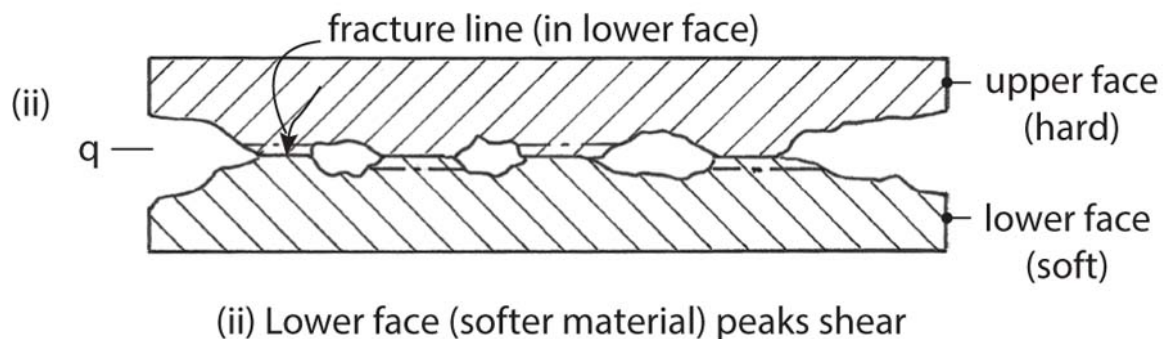
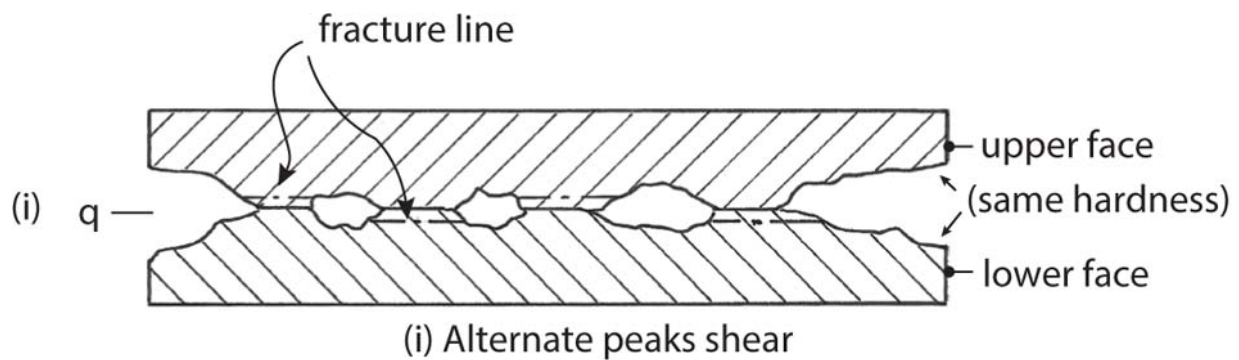


Figure 48 Surface sliding – modes of movement.

THE MARKHAM COLLIERY DISASTER

It was 6.20am on 30 July 1973, halfway down the 429m-deep shaft, as the third cage-full of miners were on their way to start the day shift at Derbyshire's largest pit. Descending normally at ~8m/s, the double cage that had been making this round trip many times every day since 1947, suddenly gathered speed and began a desperate, headlong, out-of-control fall, to smash into the pit bottom, resulting in eighteen miners killed and eleven seriously injured.

*The disaster caused great concern about this aspect of mine safety, as control of the cages was intended to be fail-safe, with great effort having gone into the design and maintenance of the winding gear in order to prevent such a cage free-fall, considered an avoidable accident unlike the intrinsically dangerous pit face work. **Photo 80** shows the general arrangement of the relevant parts of the cage winding mechanism for controlling the descent of one pair of cages with miners en-route to the pit bottom; another shaft housed a similar cage-pair ascending with miners having completed their shift, returning in a counter-balancing load to the surface. At the time of the disaster, only the descent shaft was in use, otherwise the ascending cages, which smashed into the surface winding gear, would have undoubtedly incurred even more loss of life.*

This is an important case for the extent of work undertaken, primarily by the Safety of Mines Research Establishment (SMRE), Sheffield, as presented to Parliament in April 1974. The investigation soon discovered a failure in the main loading arm (known as the 'brake arm'), which acts on the series of levers to apply, or release, the brake shoes that restrain the winding drum and hence the cable supporting the cages. This brake operates in the same manner as a brake shoe in a car drum brake, except for acting directly on the outside of the winding drum. The drum is driven and partly braked by electrical motors, applying power to lift the cages, and regenerative braking the motors to control descent, all through an electrical control system operated by the winding drum operator, the 'engine man', responsible for the safe transfer of men up and down the shafts. The drum brake is intended to be fail-safe to guard against overspeed of the cages, as it is held in the 'off' position by pneumatic power applied to the lever system. Shutting off, or loss, of this power enables the mechanical spring system to automatically operate the brake, thereby stopping the winding drum and the cages. Any failure in the control system enables the engine man's 'dead man's handle' to be pulled to activate this brake, arresting the cages.



Photo 80 Markham colliery; accident damage to pit head building.

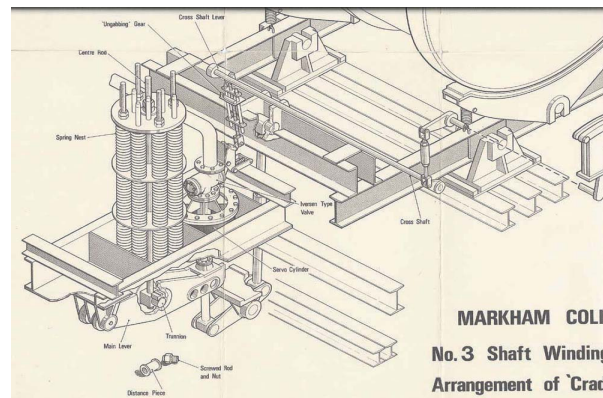


Photo 81 Markham colliery; general arrangement of drum winding mechanism.

The early discovery of the fracture in the brake rod demonstrated the sad fact that this system had been rendered useless by a failure in the critical single load-line, a fundamentally flawed system that was to result in its being finally banned in such safety-critical applications. All such arrangements were subsequently required to have back-up systems, or elements which cannot fail.

However, it was the detail of the forensic investigation into the causes of the fracture, from the design of the elements around the brake rod, condition of the material, its failure mode and metallurgical examination, undertaken by SMRE and UK universities, which provides a number of features relevant to wider application. The published report provides an excellent example of experimental work and application of current materials knowledge (see [Bibliography](#)).

The first area of analysis concerned the mechanics of the brake rod and its connection, via the trunnion, to the first brake arm. The intention of the design was to enable the brake rod to transmit the tensile load smoothly, without significant rotational resistance, to move the arm.

In practice, it was found that this bearing was deficient, with the loads being sufficient to overcome the manually applied lubrication, generating bending moments on the brake rod as it moved the brake arm.

The nest of eight double springs ([Photo 82](#)) is pre-loaded to apply a continuous tensile load in the brake rod, to which the trunnion is attached. With the brake in the 'on' position (brake drum stopped) the tensile stress in the brake rod was measured as $\sim 65 \text{ N/mm}^2$, and $\sim 75 \text{ N/mm}^2$ with the brake released, consistent with the mechanical design of the spring nest and brake lever arrangement. The investigators had to re-manufacture the brake rod, in order to experiment on the actual pit-head equipment, rather than relying solely on making models for laboratory testing. The loading on the brake rod was measured by fixing strain gauges, some 533mm up the rod from the trunnion end. These measurements showed large variations in strains, indicating a bending component in the brake rod, and making the trunnion bearing suspect. The brake rod was then transferred to the SMRE laboratory where a simulation rig enabled strain gauges to be mounted nearer to the trunnion, at 210mm from the end. Measurements with the brake rod, loaded for the 'on' and 'off' position as connected to the brake arm, showed the magnitude of stresses. These included the intentional tensile, plus the unintentional bending, stresses.

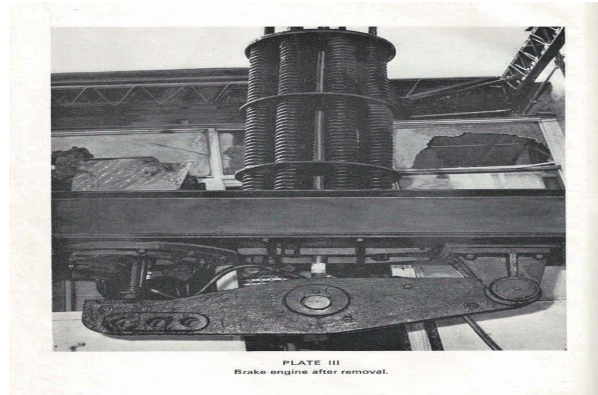


Photo 82 Markham colliery; spring nest and lever mechanism for brake rod.

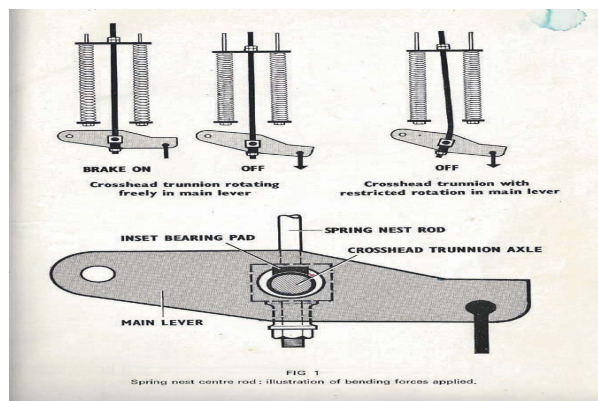


Photo 83 Markham colliery; diagram of bending moment applied to the brake rod.

The deficiency of the trunnion bearing was held responsible for causing this detrimental bending moment. With resolution of the quadrant stresses, maximum and minimum values were found, considering only the stress on the un-threaded part of the brake rod. These were found to be -75 N/mm^2 (brake off) and $+105 \text{ N/mm}^2$ (bending stress), with principal axis misaligned from the expected brake lever axis, probably due to the brake-nut seating effects. This produced, in the un-threaded part of the brake rod, a combined stress maximum of $105 + 75 = 180 \text{ N/mm}^2$.

The investigators were not able to measure the local stress at the root of the thread where failure was seen to have occurred; instead it was possible, through a series of steps, to calculate the effect of: 1) reduced section at the thread root; 2) the effect of the bending moment applied to the brake rod, by the deficient trunnion bearing; and 3) stress concentration due to the thread profile. This was the situation in 1973, but a similar case now could use 'finite element analysis' to model each of these elements, although this is still a specialized field and a classical analysis remains the most likely means for the non-corporate engineer.

- 1. The stress at the thread root is obtained by multiplying the 'on' and 'off' position stresses, of ~ 65 and 75 N/mm^2 , by the relative areas of plain rod (51.3mm dia.) and thread root (46.6mm dia.).**

So 'off' stress = $75 \times A_{\text{rod}}/A_{\text{root}}$
 $\sigma = 75 \times (51.3)^2 \times 4/(44.6)^2 \times 4$
 $\sigma = \sim 99 \text{ N/mm}^2$ – tensile

2. The effect of bending derives from the formula:

$\sigma = My/I$
 where $y = d/2$, $I = d^4/64$

From the rod and thread, where σd_R = rod diameter, and d_t = thread diameter, stress in rod = σ_R , and stress in thread = σ_t .
 As $\sigma_R = Md_R/2 I_R$, and $t = Md_t/12 I_t$, when combined, gives:

$\sigma_t = 105(51.3)^3/(44.6)^3$
 $\sigma_t = 160 \text{ N/mm}^2$ – bending

Therefore, the total stress is bending and tensile at thread core:

$\sigma_{\text{total}} \sim 160 + 99 = 259 \text{ N/mm}^2$

3. The stress concentration at the root of a thread is a function of the thread diameter, root diameter, radius at root, and any surface details at the root. However, experience shows that the repeated pattern of threads on a bolt produces a lesser stress concentration than a single thread profile in isolation. Stress concentration factors for thread roots are surprisingly controversial, and preference should be given to empirical data gathered from experience or experimentation; but for the purposes here, an analysis from Neubar (see [Bibliography](#)) provides a value of SCF ~4.3. From Peterson (see [Bibliography](#)), an appropriate SCF can range from ~2.7 to 6.7, with a value of 4.8 being possible for this case. Peterson also provides some crafty nut designs to optimize fatigue life in bolts, but assignment of fatigue strength for threaded bolts is indeed a tricky topic. Means of circumventing the problem in normal fasteners were discussed under 'fatigue prevention'.

Therefore, the local stress accumulated at the thread root is given as:

$\sigma_{td} \sigma 259 \times 4.8 \sim 1240 \text{ N/mm}^2$

This value of local stress is far higher than the measured yield stress of 346 N/mm^2 , and even tensile strength at 600 N/mm^2 ; in practice, however, this would lead to local yielding, relieving the extreme stress concentration. Beyond this there remains a sufficient factor of safety of yield/operating stress, $346/259$, FoS ~1.4, which indicates failure should not occur under non-fatigue conditions.

We now return to an examination of the material of the brake rod, having established that the loading was sufficient to prevent proper movement in the trunnion bearing,

which applied a damaging bending stress to the brake rod. Experience has shown that such a bearing should comprise a hard shaft, moving in a properly lubricated soft faced support, not the mild-steel to mild-steel combination of the trunnion shafts in the lever arm, as shown in [Photo 85](#). A series of tests with different materials and lubricants, undertaken by SMRE, proved that better material selection and maintenance would have eliminated significant bending stress and so prolonged the brake rod's life.

The investigators firstly undertook 'nondestructive testing' (NDT) of the brake rod thread around the failure region. This revealed a large number of secondary cracks in the thread root, any of which could have ultimately provided the primary crack that led to failure. This indicated that the cause of fracture was not due to an isolated material defect, and also that NDT should have been included in the maintenance regime during the twenty-one year life of the mechanism. These dangerous defects could have been detected and the brake rod safely removed from service.

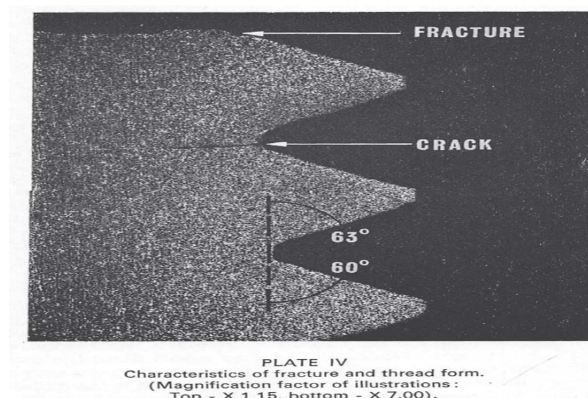


Photo 84 Markham colliery; section through the brake rod thread, showing incipient root cracking and thread form deformation, indicating high stresses.

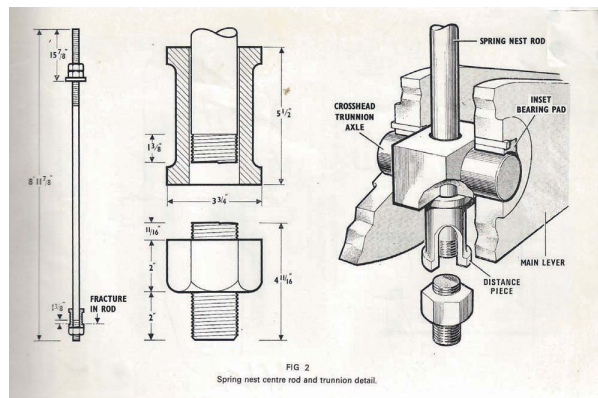


Photo 85 Markham colliery; diagram of brake rod and trunnion arrangement.



Photo 86 Markham colliery; diagram of cracks found by NDE on brake rod thread.

The next step was to re-examine the design specification, which required the brake rod to be made from the (then current) BS970, EN8 standard. The 1947 standard specified only composition limits, tensile and yield strength, and elongation, for the normalized condition. Samples taken from the rod were found to satisfy both chemistry and mechanical properties, with $\sigma_y=346 \text{ N/mm}^2$, $\sigma_u=600 \text{ N/mm}^2$, and elongation = 30 per cent – all well within specification.

However, Charpy impact tests revealed very poor impact toughness at or below room temperature (20°C), of only 10 Joules, with 100 per cent brittle fracture of the surface. Tests at 100°C provided 46 Joules and 35 per cent brittle fracture, values which would be expected for a safe application at room temperature. Metallographic examination revealed the excessive inclusions that led to this poor toughness, without having such a damaging effect on the tensile properties. Notably, the subsequent material standard for EN8, grade 080M40, made a requirement for impact properties that would have identified this problem. The microstructure also showed the clear constituents of ferrite and pearlite grains, with longitudinal alignments as expected from the manufacture of drawn bar. The absence of martensitic or similar transformation products, and the general ordered structure, proved that the material was in the specified condition of being hot worked or normalized.

However, the microstructure showed no significant flow at the thread root, indicating that it had been formed by thread cutting, rather than by rolling. As shown in [Figure 43](#), the effect of thread rolling enhances fatigue performance, as noted earlier.

The final metallurgical work involved fractography, the examination of the fracture faces with the SEM. This showed that, over the 60 per cent of slow fracture area, fatigue striations could be discerned at a spacing of $\sim 0.5\mu\text{m}$. The remaining section, around 40 per cent, showed fast failure from brittle fracture, with clear cleavage facets and no evidence of ductility, as would be expected from the impact test results at room temperature.

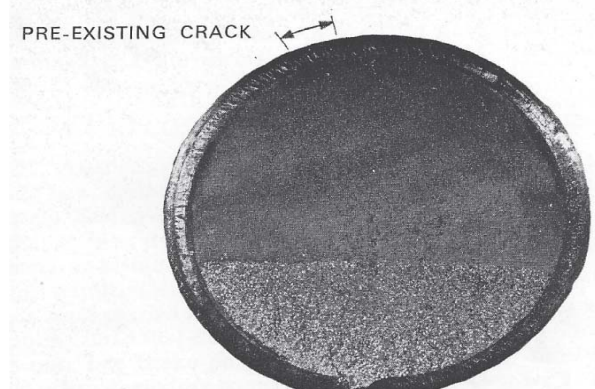


Photo 87 Markham colliery; fracture face of the brake rod, showing fatigue crack growth and single-overload final failure region.

The investigation established that the material of the brake rod met the design specification and would have proven successful provided that the loading on it was not increased by any malfunction of the lever system. However, in the inevitable presence of cyclic loading, the very low toughness makes it crack sensitive, and the cut thread provides an increased probability of initiating a fatigue crack. Whether or not this nucleates, and then grows to a crack that finally results in failure, is dependent on a fatigue analysis.

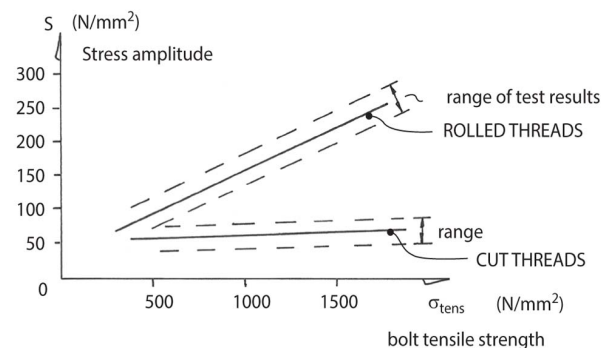


Figure 43 Fatigue behaviour – rolled and cut bolt threads.

Brake rod fracture toughness

The impact tests indicated very low toughness, and quantification of this property might provide a better understanding of critical crack lengths, which could be of future aid to NDT procedures in this type of equipment. Fracture toughness tests were undertaken at a UK university, using specimens cut from the actual brake, which naturally limited specimen sizes, but was still considered valid, given the low toughness values predicted (see [Figure 44](#)).

From the specimen size, crack length, and measured loads to failure, the fracture toughness (K_Q) was calculated as $\sim 45 \text{ MNm}^{-3/2}$, $\sigma_y = 346 \text{ MNm}^{-2}$, so $B \sim 41 \text{ mm}$. Whilst this is substantially greater than the actual specimen thickness of 22.5 mm, the test piece fractures showed only limited evidence of the shear lips associated

with plane stress (i.e. non plane strain) conditions, and the test was considered a fair approximation to a K_{1C} test. Note: $1 \text{ N/mm}^2 = 1 \text{ MNm}^{-2} = 1 \text{ MPa}$.

The stress intensity, K , can therefore be found from $K = Y\sigma_{cr}(a)^{1/2}$. From published data we have a plot of Y vs. a/D , where a is the crack length, D the thread diameter, and σ_{cr} is yield stress (see [Bibliography](#)). Taking the value of $a = 27 \text{ mm}$, the crack length at failure of the brake rod, $Y = 3.0$, and assuming $K_{1C} = 45 \text{ MNm}^{-3/2}$, we can find the critical stress required in the brake rod to initiate fracture, so:

$$K_{1C} = Y \times \sigma_{cr} (a)^{1/2}$$

$$45 = 3.0 \times \sigma_{cr} (\times 27 \times 10^{-3})^{1/2}$$

$$\sigma_{cr} = (45/3) \times (\times 27 \times 10^{-3})^{1/2}$$

$$\sigma_{cr} = 51.5 \text{ MNm}^{-2}, \text{ or } 51.5 \text{ N/mm}^2$$

Thus the stress for failure is predicted to be only $\sim 52 \text{ MNm}^{-2}$, whereas we calculated the stress applied was far higher, at $\sim 99 \text{ MNm}^{-2}$. This indicates that the actual loading condition was more complex, and it can be envisioned that the growing crack would have enabled the brake rod to deflect and thereby relieve some of the damaging bending stress, allowing a longer life than otherwise would have been the case. This is not untypical of the complexity involved in real-life situations, but the analysis does point to the low values of stress that could give serious concern for failure.

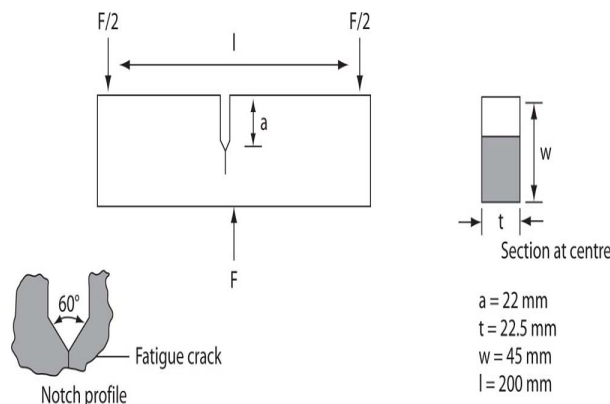


Figure 44 Fracture toughness specimen brake rod tests.

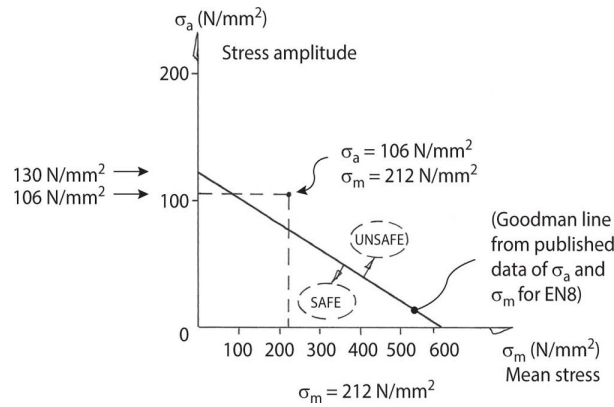


Figure 45 Goodman diagram for Markham brake rod.

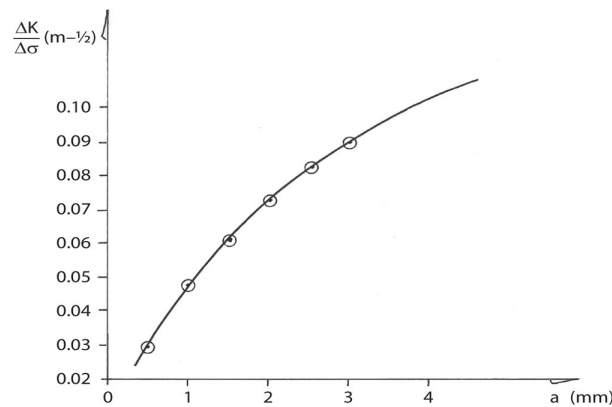


Figure 46 Standard data for EN8 crack growth calculation.

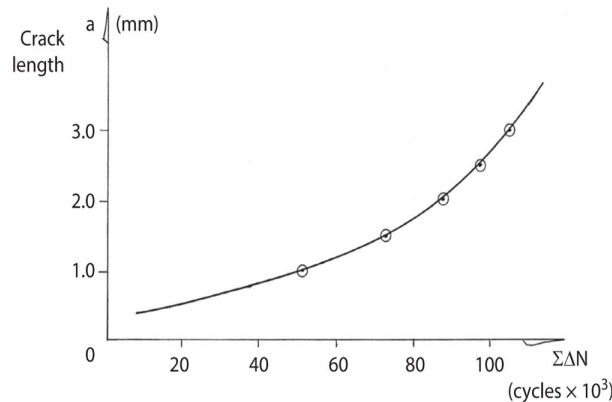


Figure 47 Estimated crack growth rate for Markham EN8 bolt rod failure.

The final piece in the jigsaw is an analysis based on fatigue. From the stress analysis outlined earlier, it is possible to determine the total effect of the loading, resolved to provide a mean stress of $\sim 130 \text{ MNm}^{-2}$, and stress amplitude as half the stress range of $\sim 212 \text{ MNm}^{-2}$, thus $\sigma_a = 106 \text{ MNm}^{-2}$.

This data can then be used with a Goodman Diagram, as described earlier, to find if these values will lead to failure (see [Figure 46](#)), based on published data for a

similar EN8 material, with $\sigma_u = 550 \text{ MNm}^{-2}$ fatigue limit $+120 \text{ MNm}^{-2}$.

From the Goodman Diagram, the mean stress and amplitude place the brake rod well into the unsafe zone, indicating a finite life. This would have required urgent attention, such as careful NDT, to check the likely areas for failure, the threaded region above the trunnion. Furthermore, the data on fatigue of steel bolts (see [Figure 43](#)) for cut threads indicates that an alternating stress of only half that calculated for the brake rod would be considered safe!

Crack growth rates, as noted earlier, can be calculated from the fracture mechanics data and plotted on da/dN vs. ΔK graphs, by setting up a table of crack length and stress intensity factors (ΔK), for each progressing crack length, as shown in [Table 30](#).

Table 30 Brake rod fatigue crack growth rate estimate.

a_o	a_c	Δa	$\Delta K/\Delta\sigma^{(m-1/2)}$	$\Delta K = 212 \Delta K/\Delta\sigma$ ($\text{Mnm}^{-3/2}$)	$da/dN \times 10^{-6}$ (mm/cycle)	ΔN (cycles)	$\Sigma \Delta N$ (cycles)
1	1.5	0.5	0.047	9.96	9.88	50.6	50.6
1.5	2	0.5	0.061	12.93	21.62	23.1	73.7
2	2.5	0.5	0.073	15.48	37.1	13.5	87.2
2.5	3	0.5	0.082	17.38	52.5	9.5	96.7
3	3.5	0.5	0.089	18.87	67.19	7.4	104.1

$\Delta K/\Delta\sigma$ from Figure 46
$da/dN = 1 \times 10^{-8} (\Delta K)^3 \text{ mm/cycle}$ from Paris-Erdogan eqn. for plain carbon steels for central steady-state crack growth
$\Delta N = \Delta a / da/dN$
a vs ΔN from Figure 47

This predicts ~100,000 cycles to reach a crack length of 2.5mm, and from estimates of the number of journeys of the cage, some 700,000 cycles were thought to have occurred before the fatal failure. As there would have been a substantial period for crack initiation, the analysis looks credible for growing the fatigue crack to its final length of ~27 mm, given likely relaxation of the bending moment as the crack lengthened.

Conclusion

A number of contributing factors were, as is often the case, responsible for the failure of the brake rod. The first recommended change was to the single load-line system, which relies on critical items but fails to safeguard against failure, either by back-up systems, or precautions such as targeted NDT.

The design and maintenance of the trunnion bearing was found wanting, creating unexpected bending loads which seriously compromised the loading in the brake rod, producing higher-than-anticipated mean and alternating stress amplitudes.

The EN8 material, installed in 1947, was found to be inferior to that expected of modern steels, with particular regard to a low value of fracture toughness, which made it crack-sensitive, a potentially dangerous state in a fatigue environment. The manufacture by thread cutting further exacerbated this sensitivity in a cyclic loading environment.

Finally, the high-cycle environment and heavy usage, combined with some fundamental material toughness deficiencies, produced the classic case of an accident waiting to happen.

Journal bearings

Bearings are fundamental to all aspects of engineering and, given the problems with surfaces in contact as noted above, means are needed to overcome these; otherwise the wheels of industry, and home workshop, would grind to a halt. The key is friction; some typical values for coefficient of friction for bearing materials given in [Table 31](#). However, these values are relatively high and would soon lead to problems unless a suitable lubrication system is employed.

Table 31 Friction coefficients for some typical bearing alloys.

Material	H_v	~ coefficient of friction, sliding, no lubrication, dry air
Cu – Pb	35–55	0.2
Phosphor Bronze	85–120	0.35
Si – Al	50–60	0.5
white metal (Pb base)	18–22	0.6–0.7
white metal (Sn base)	20–24	0.6–0.8

The theory of full-film lubrication is shown in [Figure 49](#), where a wedge of oil is drawn around the shaft, supporting it within the bearing. The choice of lubricant enables its shear stress, derived from the viscosity, to be balanced against the imposed loads, minimizing frictional losses whilst maintaining the shaft out of direct contact with the bearing. Over-high viscosity produces excessive

'drag' on the shaft; insufficient viscosity would allow the shaft to make metal-to-metal contact with the bearing, where at low speeds the lubrication wedge may not be available. Careful choice of bearing material is used to overcome the transient conditions that could otherwise lead to rapid adhesion wear, such as in starting a car engine from cold.

Plain bearing materials have been described in [Chapter 2](#), and require: low enough hardness to allow conformance to the mating part; being sufficiently inert to avoid corrosion; a low enough melting point for local plastic flow in cases of temporary lubrication loss; and surface softness sufficient to embed particles entrained in the lubricant. Typical of this class is the white metal lined, shell-backed, big-end bearings of cars, an extreme case of loading with near-zero wear rate, possible with modern materials, design and lubrication.

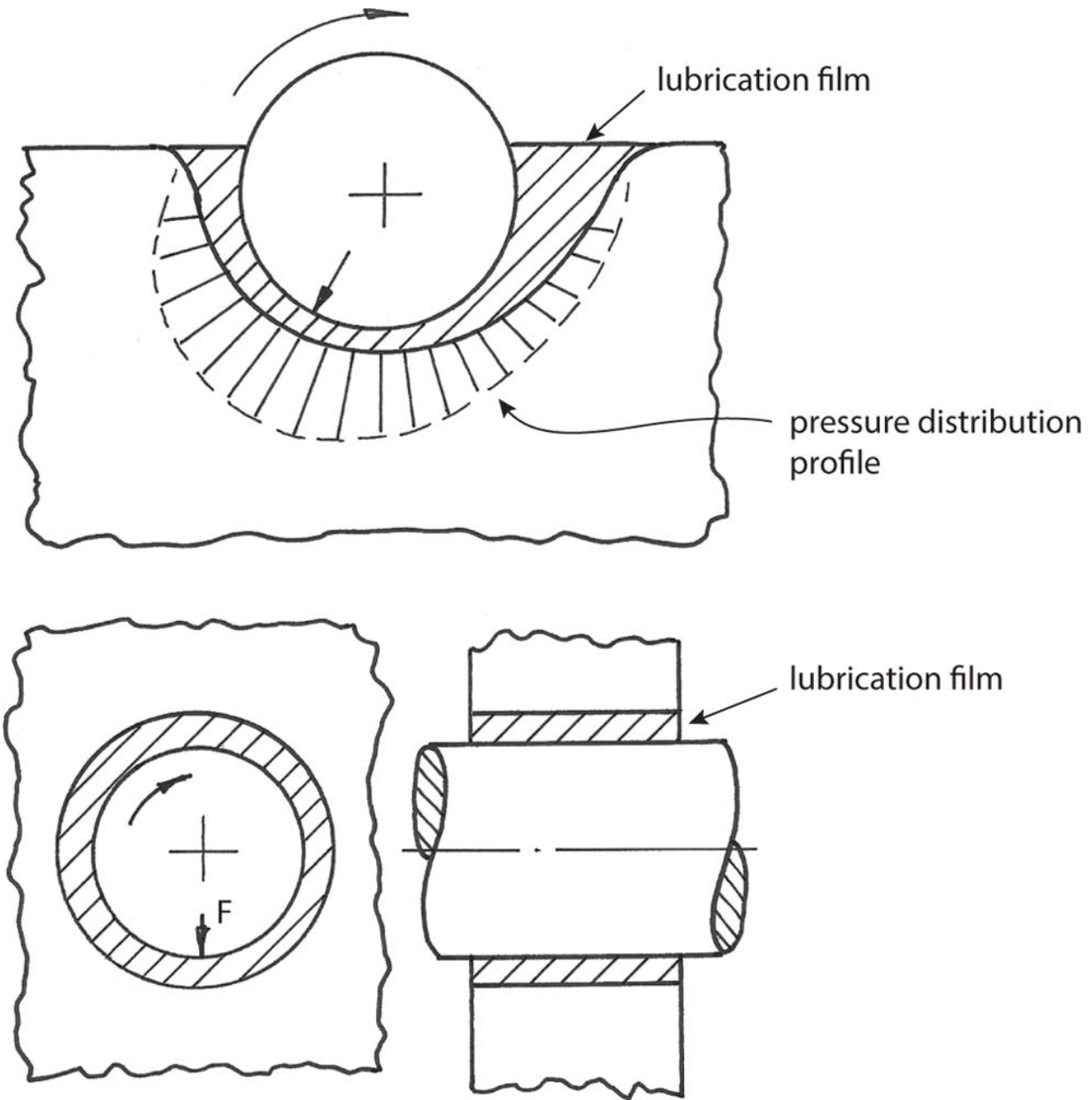


Figure 49 Lubrication of shafts.

Surface films and corrosion

The metallurgical aspects of surfaces and their various films is a highly complex subject which plays a critical role in wear and adhesion. This operates at layer thicknesses as low as one nm ($1 \times 10^{-9}\text{m}$).

The behaviour of surfaces, in the absence of lubrication, is strongly dependent on the operation of surface films, and can subject wear to stages: low wear rates are produced at loads below those where the surface films remain intact; high rates arise upon their disruption; and then lower rates resume as factors such as changes to near-surface structures kick in. A common feature with oscillating parts is fretting damage, where the oxide films are disturbed and may lead to film fracture and dispersal, through fretting corrosion. Chemical action can result in damage to oxide films, where repeated loss and re-formation can lead to wear particles which form on the surface and contribute to an abrasive wear compound.

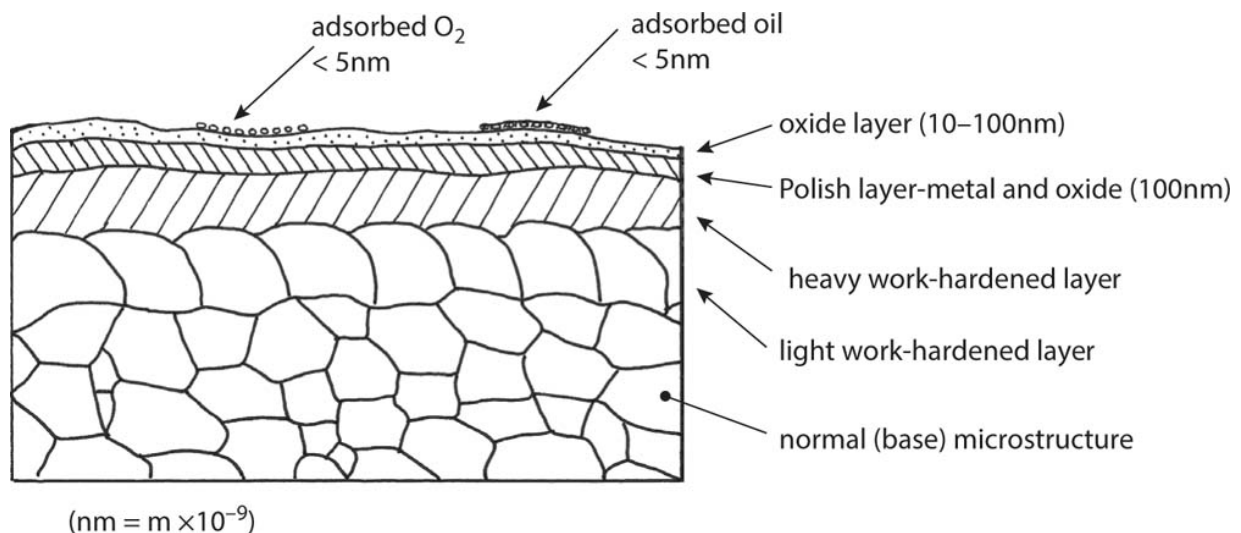


Figure 50 Metal surface films and microstructure typical of polished surfaces.

Abrasion wear

This is a wear mechanism that occurs when a hard material is drawn over a softer and the former's roughness causes a gouging of the latter (like a file being drawn across a block of mild steel). This can occur when hard particles are not embedded but present cutting edges, leading to high wear rates. Where the softer material is able to embed the hard particles sufficiently, then they cause far less damage and may even be made sufficiently safe to have no further

impact on wear. Abrasion wear is often associated with the production of small-particle wear debris, in dry-surface movement.

Fatigue wear

This is the common mechanism of wear experienced in rolling bearings, with the ball or rolling element subjecting the bearing race to high frequency compressive cyclic loading. Despite the normal reluctance of metals to suffer fatigue failure under compressive loading, the roller bearing is a special case due to the extremely high local stresses, high strength materials, and large number of loading cycles seen by the bearing. The small contact area of the rolling elements provides surprisingly high stresses, and this can ultimately lead to a spalling failure of material, from an originating crack near, but often below, the surface of the bearing race. Once established, the crack can grow until a piece of material is ejected from the surface, creating a deteriorating pit, much as a pothole in the road is created and enlarged by the attack of repeated vehicle-wheel loadings.

A similar mechanism can also occur in sliding bearings, even thin wall bearing shells, where sufficient cyclic loading occurs, although this is normally precluded by the much lower stresses due to the larger contact areas. More seriously, it is possible for 'delamination wear' to occur where material is broken off, causing severe work-hardening of the surface layers and making for crack sensitivity. Repeated loading can lead to fatigue cracks and spalls, typically up to 100µm diameter.

Wear prediction and prevention

There are no simple rules to determine the wear of metals, but general guidelines can be derived from the mechanisms of adhesion, abrasion, delamination, fatigue and corrosion. Generally, harder materials wear less, but it is the combination of metals and lubrication that is critical. The well-known propensity for stainless-

steel to stainless-steel contact to 'gall' is an example of the inter-atomic forces associated with the tenacious oxide film, also responsible for their corrosion resistance. Conversely, cast irons can readily run together, due to the large amounts of free graphite, which provides a self-lubricating facility.

An analysis of wear can involve an examination of wear debris, surface appearance, and metallographic section, to reveal oxide layers and near-surface microstructure. From these evaluations, the mode of wear can be established and appropriate action taken.

The best method of avoiding wear is to physically separate the surfaces, usually by a lubricating film. Depending on the severity of loading and operating conditions, the lubrication system can be chosen, from pressure-fed oiling, grease servicing, or occasional manual oiling. Where this is not possible, or is likely to encounter intermittent failure, resort can be made to materials like grey-flake cast iron, or oil-impregnated sintered bushes. This was an important factor in the trunnion bearing in the Markham Colliery disaster.

Corrosion

Despite the importance of corrosion, with estimates of its cost to the UK economy of £5–10 billion per annum, and its obvious impact in the home workshop, it is such an expansive subject that it can only be touched upon here, with further information listed in the [Bibliography](#).

Surface films

A topic mentioned in 'wear' was the effect of surfaces and the oxide films always present on metals. Whilst this is a highly technical subject on a truly microscopic scale, there are some features that support at least a basic understanding. Oxide films have a profound impact on corrosion, joining, and even fracture in high-temperature-applications like turbine blades.

Possibly the greatest limitation of mild steel is its poor corrosion resistance, in environments other than those free from the moisture experienced in everyday use in most parts of the world. This has been well known since ancient times, and protective coatings ('boundary protection'), such as painting, are taken for granted. The reason for the familiar rusting of plain and low-alloy steels is the porous oxide film that enables the corrosion reaction to progress by its failure to isolate the base metal from the environment. The cases will be all too familiar to those of a certain age who experienced the dramatic demise of their monocoque design car bodies, as introduced in the 1950s, within a few short years. This design enabled an aggressive wet environment, often trapped inaccessibly behind wings and door sills, in the absence of the surface protection so carefully applied to the visible outer skin, to corrode at an alarming rate. Whilst car roofs never rusted, the lower parts disintegrated, to the delight of the MOT tester! The quite remarkable transformation in corrosion resistance of cars since the late 1990s has been not only a triumph of technology, but even more the realization that obsolescence is better prompted by design innovation than product disintegration...

The basis of this improvement is principally the extensive and highly developed use of 100 per cent coverage of the steels, particularly with a galvanizing, zinc-based, treatment and further thin protective coatings. Zinc is particularly desirable, providing an anodic protection mechanism. Any break in the coating is self-healing, due to the difference of emf in the Galvanic Series (see [Table 32](#)), which produces preferential attack of the coating (anode), instead of the steel substrate (cathode). Just how effective this process is can be seen in the oft-neglected garden items such as galvanized steel dustbins and wheelbarrows, which remain un-corroded despite long-term exposure to damp conditions. Even farmers' gates and stiles will remain intact until eventually the coating is worn away with the passage of time.

The understanding of the science of surface films is only a twentieth-century development, although it has been intuitively understood through the ages, for instance in the famous Delhi Pillar,

which has survived in pristine condition despite the most humid environment (it is made of cast iron endowed with ~6 per cent silicon, producing an efficient protective SiO_2 film). And it has only been a hundred years since stainless steel was invented in 1912 and patented by the Krupp engineers Benno Strauss and Edward Maurer, and before the world's first stainless steel cutlery was produced in the Portland Works, Sheffield, UK, in 1913. This provides an interesting context when discussing the storage, in stainless steel cans, of nuclear waste, which requires corrosion resistance safety for 10,000 years.

It was found that a minimum of 10.5 per cent Cr added to steel would render it startlingly corrosion resistant to the common environments. This has led to a wide range of applications, from the iconic Chrysler Building in New York and the Thames Barrier in London, to our domestic cutlery. It should be noted, however, that stainless steel can be highly susceptible to localized corrosion attack, such as pitting corrosion or stress corrosion cracking (SCC) in certain environments, particularly those containing chlorides. This has led to the development of special alloys such as the Duplex stainless steels, but there are still instances where they fare less well than plain C steels – an illustration, if nothing else, of the complexity of corrosion engineering, where one must always be wary of generalizations. First-hand experience is as priceless as is the medical consultant's advice as compared to the layman's self-diagnosis from the Internet.

Mild steel readily forms the oxide FeO_2 , which, being porous, does not effect a complete isolation from the environment, allowing the nucleation of a chemical cell to drive the familiar rusting. The nature of the oxide film enables the continuation of the process, rather than a blocking, which occurs in the corrosion-resistant systems. The oxide films Cr_2O_3 , Al_2O_3 and SiO_2 are examples of those offering protection to further attack and consequently protection from corrosion; Cr_2O_3 is the agent of protection for stainless steel, and Al_2O_3 the film that naturally forms on aluminium, conferring its characteristic immunity. The value of these natural

films is that they are self-generating, rapidly forming each time a new surface is created, such as in machining or abrasion.

The downside to these tenacious oxide films is that they inhibit joining, as seen when attempting to solder stainless steel, only possible with highly aggressive fluxes and not normally a practical proposition. The aluminium oxide, Al_2O_3 , has the additional problem of a higher melting point than the base metal, requiring particular action to dissipate before fusion can proceed. This is only achievable with special fluxes for gas welding, or the use of AC power to agitate the molten pool in TIG welding.

Wet corrosion

The most familiar occurrences are general corrosion, pitting corrosion, stress corrosion, galvanic corrosion, and corrosion fatigue.

General corrosion implies a relatively consistent loss of metal over the whole area of exposure to the environment, often perfectly manageable where there is sufficient material and particular weaknesses do not arise unexpectedly. The great economic benefits of using mild steel means that it is often practical to accept replacement, rather than select a more expensive material with superior corrosion resistance. An example is the car exhaust silencer box, where it is not practical to effect full protection should it be fabricated from the usual mild steel. Surface treatments are destroyed by the welding process, and re-treatment difficult to achieve on the inside, without disturbance of the internal sound attenuation material. The best option is the use of an inherently corrosion-resistant material, such as an austenitic stainless steel (e.g. 316L), but this is not without an economic impact.

Table 32 Galvanic series – relative potentials (emf) of metals in seawater or similar.

0.3	0.2	0.1	0	-0.1	-0.2	-0.3	-0.4	-0.5	-0.6	-0.7	-0.8	-0.9	-1	-1.1	-1.2	-1.3	-1.4	-1.5	-1.5	-1.6	
Least noble/corroded (ANODIC)																					Material
																					magnesium
																					zinc
																					beryllium
																					aluminium alloys
																					mild steel
																					low alloy steels
																					aluminium bronze
																					naval brass
																					tin
																					copper
																					admiralty brass
																					stainless steels – martensitic
																					90-10 copper-nickel
																					lead
																					silver
																					stainless steels – austenitic
																					titanium
																					gold
																					platinum
																					graphite
Most noble/un-corroded (CATHODIC)																					

Note that stainless steel welding has the potential to suffer from ‘weld decay’, where the HAZ reaction can cause the precipitation of Cr atoms, combining with C to form Cr_{23}C_6 carbides, so depleting a region adjacent to the weld of Cr, with consequent loss of corrosion resistance. However, this has long been understood and all competent fabricators avoid this effect by using only the low-C grade of stainless, such as 316L. Stabilization agents, such as niobium (Nb) are also used in the electrodes, and this is no longer a problem unless a higher grade of stainless steel, with C > 0.04 per cent, is used.

Pitting corrosion is a phenomenon that can be a problem in normally corrosion-resistant materials, such as stainless steels or

copper-base alloys. The mechanism is complex in detail, but essentially provides a concentrated attack where a pit is generated from a highly localized corrosion cell that fails to be self-healing, as is normally the case. Crevice corrosion is a form of pitting corrosion that can afflict stainless steels in particular environments, where small gaps in mating parts like vessel flanges are encountered. Designs should therefore always avoid crevices in stainless steels wherever possible, in cases of potential problems. Once again, for-critical projects expert advice is required.

Stress corrosion is an insidious form of attack that can render the corrosion-resistant alloys unusable, but requires stress to drive a transgranular crack across a section exposed to particular environments. Examples include 'season cracking' in brass, noted earlier (see [Photo 58](#)) and stainless steel in chloride environments. The stress may be provided from external loading, but the hidden residual stress, such as from fabrication, can be sufficient to eventually destroy the product. The best precaution may be to avoid internal stresses, by heat treatment if possible.

Galvanic corrosion is always a potential problem when metals from different positions in the Galvanic Series are coupled together, as in a steel bolt holding aluminium plates together. Corrosion is not inevitable and often simple precautions like insulating faces with a gasket may be sufficient, or the rate of corrosion may simply be acceptable. The effect on corrosion rate is often of paramount importance in practice, and evaluation by experience is the most reliable guide to success within budget.

Corrosion fatigue, i.e. when corrosion occurs in a fatigue environment, is particularly problematic, the effect usually far greater than the sum of the individual rates of corrosion and fatigue on crack initiation and growth. Perhaps rather like a walk in the mountains with both wind and rain being sufficient to thwart the most ardent adventurer, either the corrosion agent needs arresting, the operation modified, or service life be curtailed accordingly.

5 *Back to basics*

This chapter could prove handy for those lost souls, like myself, who are inclined to read books by flicking through from the back, rather than beginning diligently from the front. It aims to provide a rather more basic understanding of the subject of ‘metals in the service of man (in the shed)’. In order to tackle these somewhat tricky aspects in a potentially useful manner some key topics are selected. These form the basic tools, as deployed by the materials specialist, to see beyond the superficial to an understanding that should be rewarding in its own right and may even be of practical use.

THE STRUCTURE OF METALS

Whilst engineering artefacts can be of truly massive scale, as in the elements of bridges, buildings or buses, the nature of metals is determined at the small and, more commonly, the extremely small scale. The former is classified as ‘macroscopic’, where effects may be discerned by the unaided eye, perhaps with some degree of special preparation. Fracture faces can provide some valuable information to the naked eye of the experienced investigator, such as revealing the macroscopic ‘beach marks’ often seen on fatigue failures of rotating shafts; surface cracking or distortion of an overloaded steel arm; the grossly oversized grain structure of a brass casting polished by repeated handling; or a section cut through a large weldment when polished and etched.

But beyond the macroscopic, the essence of metals requires a ‘microscopic’ examination in order to show, for instance, the usual grain structure. This is typically revealed at magnifications of ~100 to 500 times in the optical microscope. In a similar manner, fracture

faces usually require hundreds- or thousands-times magnification, to reveal fully the critical detail that can be used to recognize the profiles of single-overload or fatigue failure, enabling the forensic metallurgist to determine if an aircraft explosion was caused by a bomb or slow crack growth, fatigue, or structural failure. The study of fracture faces, and the invaluable insight they can provide, has been completely transformed since the [introduction](#) of the scanning electron microscope (SEM) in the mid-twentieth centuries, enabling up to ~10,000-times magnifications. Previously it was impossible to achieve the optical depth of field at greater than the macro scale, sufficient to keep these rough surfaces in focus. The SEM effectively provides a 3D image, and enables resolution of extremely small features. The optical microscope, by contrast, has a very small depth of field at the required magnifications, and can therefore only view flat surfaces, although this is sufficient for viewing polished and etched specimens to reveal basic microstructures.

There are, however, many important features that were still hidden beyond the optical microscopy of the nineteenth and early twentieth centuries. Then, in the latter half of the twentieth century, the transmission electron microscope (TEM) raised the game, with special specimen preparation methods, from the previous norm of 1,000-times magnification to 100,000-times magnification. This began to reveal the world of sub-microscopic structures, and confirmed theories such as the dislocation behaviour described later. There followed several developments in electron microscopy, such as the 'scanning tunnelling electron microscope' (STEM) and its derivatives, which have now virtually reached the holy grail of resolving direct images down to the basic atomic structure of crystals. At around 10^{-9} to 10^{-10} m, this is the realm of the 'nanometre' and beyond, technology well-established in the electronics field, with much to offer in the further study of metals. So, despite mankind's long-term association with metals there remains a remarkable amount still to glean at the fundamental, and sometimes even macro, level.

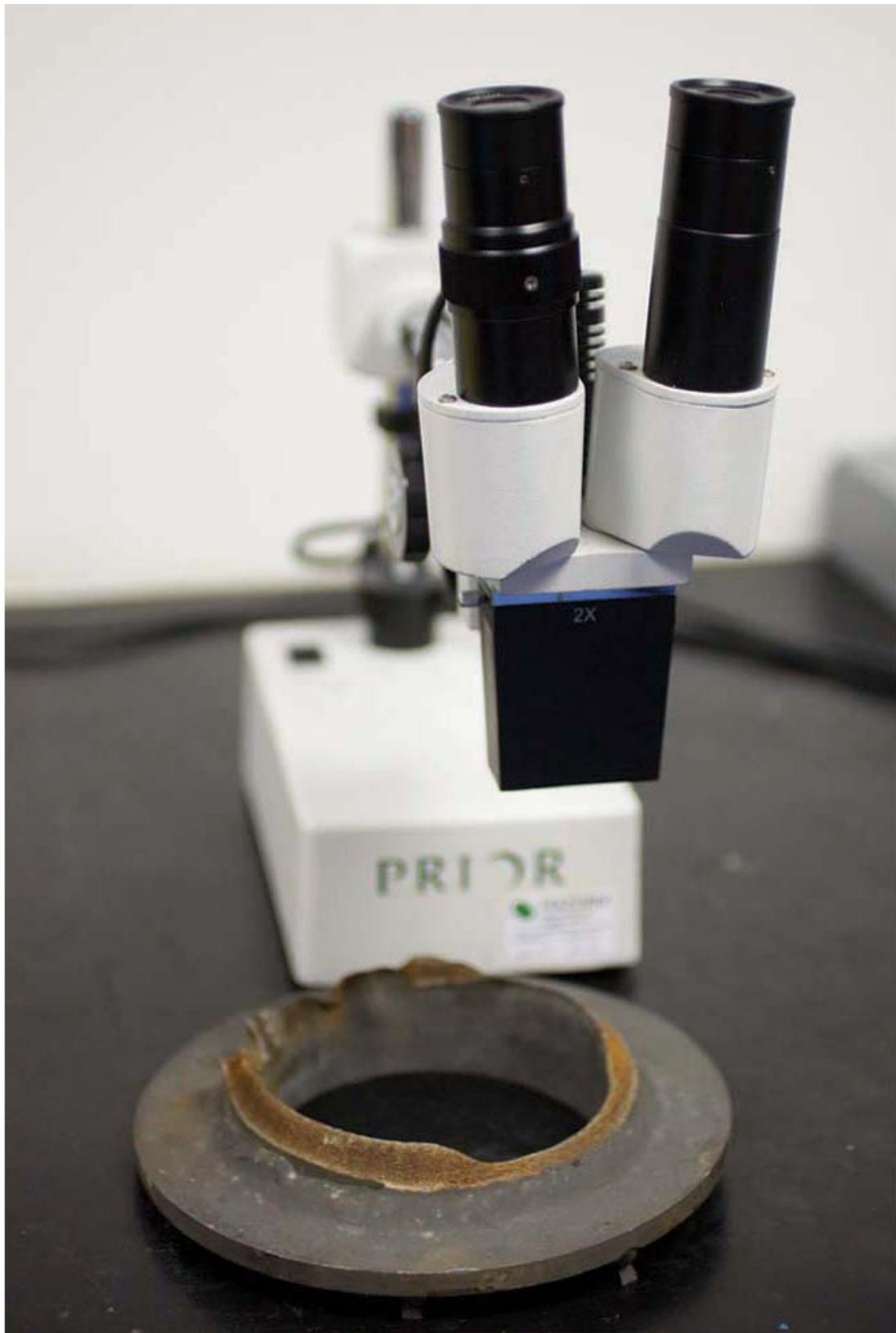


Photo 88 Low power magnification for optical examination of macro specimens.



Photo 89 Optical microscope for metallographic micro testpiece examination.

PREPARATION OF SPECIMENS – MICROSCOPY AND FRACTOGRAPHY

Macro specimens

Typical applications of a macro sample include preparation of a section through a joint, such as a butt weld in a plate, or a fillet welded 'Tee' piece. A section is cut, usually by saw, and the section face polished, through a series of increasingly finer grades of abrasive paper, until it is possible to transfer to fine polishing methods. For industrial purposes, commercial test houses, such as MIS Ltd, use diamond paste impregnated polishing wheels on which to achieve as near a scratch-free surface as the project demands (see [Photo 67](#)). An etchant, such as 10 per cent nital (nitric acid in water) for steel, is carefully applied in order to reveal the macro structure to the naked eye, or low power (<20-times) magnification. This is often quite sufficient for illustrating the individual weld beads, and defects such as lack of 'root fusion' or 'side-wall fusion'. Even a simple cut-and-ground specimen is adequate to check the integrity of a fillet weld, which it is normally impossible to determine from an NDT check. This is a useful technique, but obviously relies on preparing a spare sample for a destructive test.

Micro specimens

The first resort is preparation for viewing in the optical microscope, as our usual alloys have a grain size and features such as transformation products, like pearlite or martensite in steels, that require at least 100-times magnification to resolve, although normal optical microscopy is limited to 500- to 1000-times magnification.

The methods are as described for macro preparation, but a much smaller specimen is used, a typical area of examination amounting to only a few mm². This usually means cutting the sample into a piece too small to handle safely in the grinding and polishing process, and therefore it is usually set in a plastic mould, traditionally with a thermoset compound. This can then be more easily handled as the process needs to be finished with a virtually perfect scratch-free polish, so as not to interfere with the final detail of the surface topography. When etched, this reveals the arrangement of grains, grain boundaries, and the transformation products. These can show as-cast structures, the effect of heat treatment, or processing such as rolling of wrought material. The effects of corrosion, such as stress corrosion cracking in brass or stainless steel, and the heat treatment of steels with through-hardening or case-hardening is readily examined by this method. Microstructures of cast irons immediately show whether a supposed ductile, spheroidal graphite, iron has developed the requisite spherodized graphite shape, or retains the long-flake graphite of the untreated cast iron.



Photo 90 Water lubricated abrasive grinding disc for initial specimen preparation.



Photo 91 Commercial test house specimens for weld procedure testing.

Although a fundamentally simple process, as developed by Henry Clifton Sorby in Sheffield, UK, who in 1863 became the father of modern microscopy, it does demand equipment not normally found in the amateur's shed and so usually resort is made to a commercial test house (see [Appendix II: Resources](#)). However, it can be useful to know the general principles and the benefits, if a serious task is to be undertaken where verification is required of material processing, such as heat treatment.

In the case of a real-life failure investigation the above would be supplemented with mechanical testing, such as tensile and impact tests, and chemical analysis. This would establish whether the material matched the design specification, as illustrated in the Markham Colliery accident investigation, before a more general engineering analysis is progressed.



Photo 92 Macro viewer, Brinell and Rockwell laboratory hardness testing machines.



Photo 93 Traditional floor-standing manual Vickers hardness tester.

Beyond optical microscopy, the methods of preparation and examination in the electron microscope do not concern us here, except the more detailed (and very expensive!) fractographic study of fracture surfaces. Preparation often involves only simple cutting of a sample piece, carefully preserving the surface from damage, of around 12 mm cube. This is then simply cleaned and put into the SEM vacuum chamber for examination.

One final technique, relatively recently made commercially available, albeit at an industrial cost, is the use of special acetate replicas which permit the viewing of surface microstructure as a non-destructive process, rather than the normal method of cutting out a sample, which often destroys the article under examination. This method uses a special portable polishing kit to prepare a very small area on the material surface by grinding, polishing and etching a small spot, onto which is applied the plastic replica material. When set, this is removed with its faithful replication of the surface microstructure. This may be viewed in the SEM to provide a fine view of the microstructure, just as effectively as described previously with the optical microscopy of samples cut from the job.



Photo 94 Detail of Vickers hardness tester, traversing table moved to measurement position.

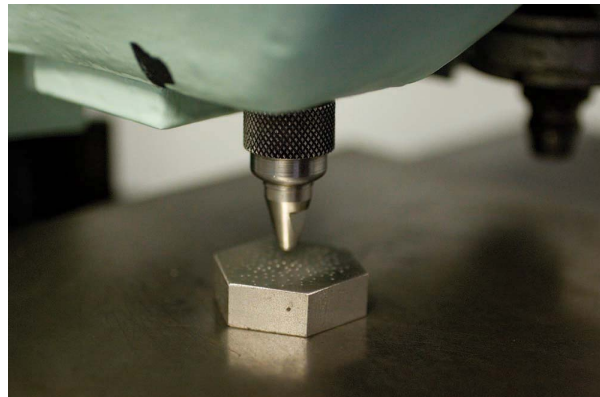


Photo 95 Vickers pyramid indentation of calibration block, for checking accuracy of measurement.

Crystallography

Whilst crystallography was the subject of a long-ago lecture course that truly struck horror into the hearts of us poor, baffled students, as delivered by an eccentric (to put it kindly) lecturer who was under the remarkably mistaken impression that he was addressing a class including Messrs. Crick and Watson – this is a topic a *little* of which is a good thing.

Solid materials exist in either an amorphous or crystalline state. The former do not directly concern us here, being familiar in many plastics with their loose and irregular – amorphous – arrangement of atoms; this means, essentially, that when deformed they don't

exhibit elasticity but change shape plastically as their name indicates, at least for the thermoplastics.

Pure metals, on the other hand, are defined by their regular crystallographic nature, where each particular metallic element or phase comprises a strict arrangement of atoms in, as it turns out in practice, almost exclusively only three different atomic cellular structures: body-centred cubic (bcc), face-centred cubic (fcc), and hexagonal close-packed (hcp), as shown in [Table 33](#).

(The only exception in the common metals is tin (Sn), which has a body-centred tetragonal structure).



Photo 96 Model of bcc (body centred cubic) atomic structure.



Photo 97 Model of fcc (face centred cubic) atomic structure.



Photo 98 Model of hcp (hexagonal close packed) atomic structure.

As can be seen from the case of iron, certain metals can exist in different 'phases', the name given to structures of different atomic arrangements, crystallographic structures, at different temperatures. This feature of different phases of the same metal is known as 'allotropy', and is of vital importance in the understanding of heat treatment and deformation. Many metallic elements and alloys exhibit this facility, of particular importance for the iron-carbon system from mild to the low alloy steels. Allotropy featured when the tin in the coat buttons of Napoleon's army changed structure to the brittle, low temperature phase, during the horrors of the Russian winter campaign. Although this was presumably amongst the least of their problems, it was of dire consequence to Captain Scott's South

Polar Expedition, when a similar fate befell the tin-alloy solder of their paraffin tins, causing catastrophic loss of the precious fuel.

Table 33 Crystal structures of pure metals.

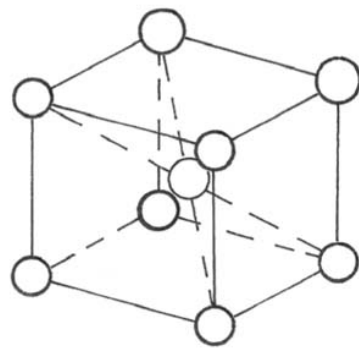
bcc	fcc	hcp
Chromium (Cr)	Aluminium (Al)	Beryllium (Be) *
Molybdenum (Mo)	Copper (Cu)	Cadmium (Cd)
Niobium (Nb)	Lead (Pb)	Magnesium (Mg)
Tungsten (W)	Iron (910 – 1400°C) (Fe)	Zinc (Zn)
Iron <910°C; >1400°C (Fe)	Nickel (Ni)	Cobalt (Co)
Vanadium (V)	Silver (Ag)	Titanium (Ti) *
	Gold (Au)	Zirconium (Zr) *
	Platinum (Pt)	

* indicates also allotropic.

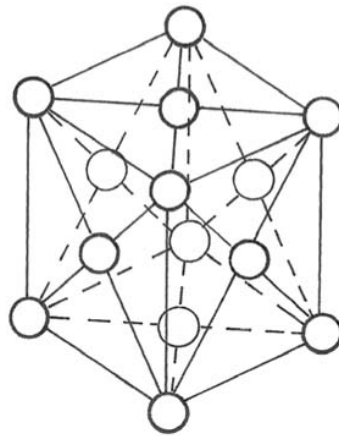
The crystal structures of [Table 33](#), as illustrated in [Figure 51](#), can conveniently be represented by atoms as billiard balls, which always fall into this crystallographic arrangement as the smallest building block of concern here. This structure is useful in understanding how pure metals deform *plastically*, as opposed to *elastically*, where in the latter case the atomic bonds between the atoms, and throughout the higgledy piggledy, ‘polymorphic’, microstructure of real metals, are simply stretched – but not beyond the point at which they will no longer spring back, elastically, to their original position.

In plastic deformation, as described later, dislocation and slip theory shows how pure metals can change shape, deform, without fracture, by a controlled movement across particular atomic planes of close-packing of the atoms. Thus, bcc structures slip along [110] planes, fcc along [111] planes, and hcp along [0001] ('basal') planes. These are simply planes through the basic crystal structure shown below. We need pursue no further the somewhat esoteric geometrical world beyond these 'Miller Indices', which identify these slip planes, into the stereographic projection and the like, primarily of interest to the research scientist engaged in the examination of alloy phases and their complex transformations.

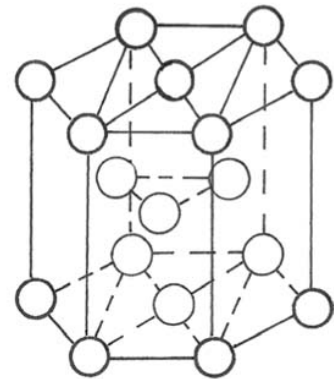
These crystals produce a 'space lattice', repeated precisely throughout three dimensions until interrupted by a 'grain boundary', abutting the next grain which is orientated, in all cases bar the 'single crystal', at a different angle of growth. Thus exists the microstructure as seen in the optical microscope, albeit only across a two-dimensional plane. These tiny aggregates of grains produce an arrangement akin to a three-dimensional jigsaw puzzle, but with the individual pieces held together by the strong 'metallic bond' that exists across normal grain boundaries. Grain boundaries are therefore not regions of weakness – unless degraded, by such as attack in the case of intergranular corrosion. Being only a few atoms wide, they are beneficial to the microstructure, as smaller grains are normally associated with higher strength and toughness. According to the 'Hall–Petch' relationship, yield strength (σ_y) is proportional to the inverse of the square root of the grain diameter; typically, $\sigma_y \propto D^{1/2}$, for steels.



(i)
Body centred cubic
(bcc)



(ii)
Face centred cubic
(fcc)



(iii)
Hexagonal
close - packed
(hcp)

Figure 51 Principal crystal structures, pure metals.

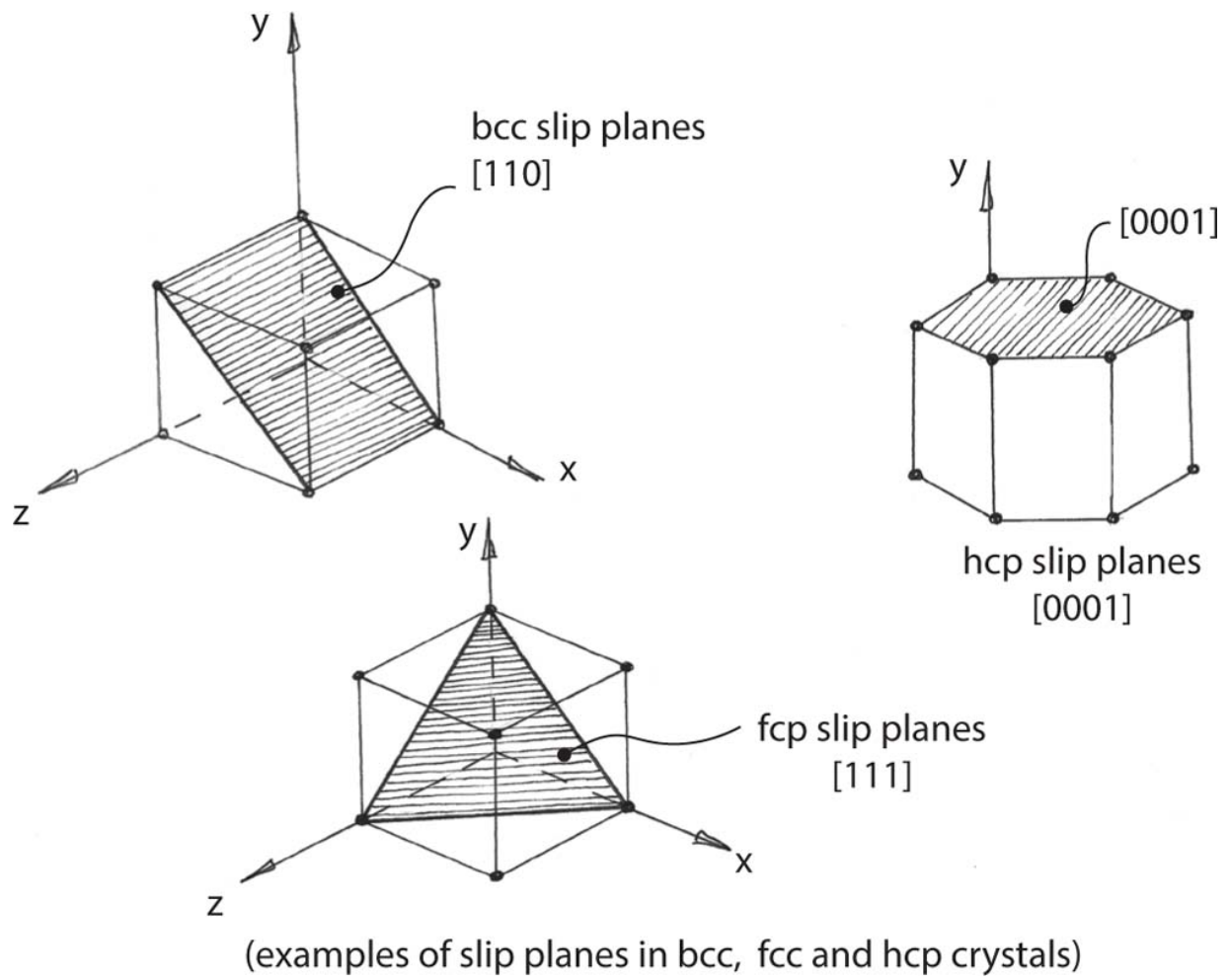


Figure 52 Common slip planes, basic crystal structures.

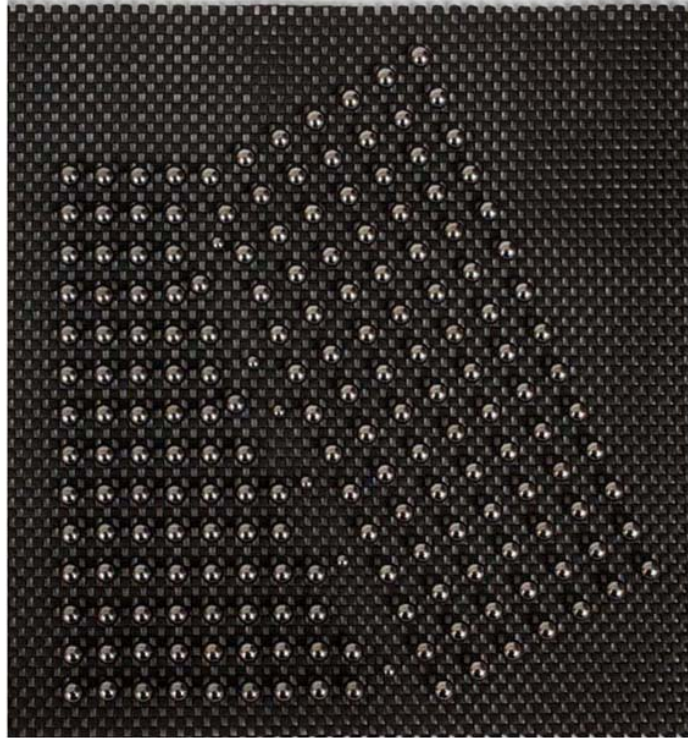


Photo 99 Ball bearing model of metallic grains abutting at a grain boundary.

The primary virtue of reducing grain size is to provide barriers to dislocation movement, so that slip is contained over shorter and shorter distances as the number of grains increases and their mean diameter reduces. This is the case for 'grain boundary strengthening' which increases the yield strength by reducing grain size and therefore increasing dislocation pile-up; but not to infinity as a limit to yield strength is observed around 10 nm (nanometres), where it is found that grain boundary sliding can occur, limiting further strength increases. The actual relationship between grain size and strength varies in different metals but the pattern, as found independently by Hall and Petch in the UK around 1954, is now well established.

During heat treatment, increasing temperature provides the 'free energy' that tends to expand the grains, just as soap bubbles tend to grow, with larger ones consuming the smaller in the effort to reach the lowest 'free energy' of a single bubble. This 'grain growth' can occur during prolonged heating at high temperature and must be avoided, or countered by the 're-crystallization' process described

later in this chapter, otherwise the mechanical properties of the material may be significantly impaired. Such might occur during extensive work in the blacksmith's forge, with repeated heating, but is of little consequence for low-stress applications such as decorative ironwork. This would not be the case for hard-working engineering components such as valve springs or gearbox internals.

The case of the single crystal was of more importance to the experimental scientist examining deformation behaviour at a fundamental level and hardly of practical application – until it was adopted as the accepted method upon which modern high-duty turbine blades are based. This is a special case where the design failure criterion is long-service creep, the nickel alloy blades being designed for the highest possible operating temperatures at which grain boundaries were found to facilitate creep. Their elimination has enabled a further step towards the ultimate metallic material for this application, as noted earlier.



Photo 100 Ball bearing model of a 'vacancy' (missing) in a metal grain.



Photo 101 Ball bearing model of crystal array, containing two 'interstitials', (superimposed, smaller, atoms).

The above has dealt primarily with pure metals, and in the practical world of engineering materials concern is with alloys produced to meet the familiar demands discussed throughout this book. The alloying of pure metals has a profound influence on the structures discussed above, such as for steels, an alloy of iron and carbon (Fe-C) plus a number of other elements like manganese and sulphur as minor additions or residuals, is seen even in a 'simple' plain carbon steel. Juggling the element balance, principally carbon which has a particularly profound effect in steels, leads to concepts of hardenability and transformation in the basic crystal structures. Such a transformation product is martensite, which holds the key to understanding how and why such effects can be produced. In the next section we continue a little further on the basis of pure metals with the concept of dislocations, to understand why real metals perform as observed.

Dislocations

From the atomic structures discussed earlier, the theoretical strength of a metal can be estimated from the inter-atomic bond strength. Similar calculations can be made for other materials, such as ceramics, plastics and composites. It can be shown (see [Bibliography](#)) that an ideal yield strength, σ_y , can be derived from the atomic spacing in materials to give σ_y of between $E/8$ to $E/15$; i.e. $E/15 < \sigma_y < E/8$, where E is Young's Modulus. It is apparent that the ceramics and plastics can approach the theoretical strength of $\sigma_y/E \sim 0.1$, whereas the pure metals reach only $\sigma_y/E \sim 0.0001$ (a thousand times inferior); commercial alloys $\sigma_y/E \sim 0.001$ (a hundred times inferior); and even the very best, fully strengthened, alloys can only approach $\sigma_y/E \sim 0.01$ (ten times inferior).

This nagging discrepancy between theoretical and observed strength in metal systems was eventually reconciled in 1934 by the work of Taylor, Orowan, Polanyi and Burgers, who postulated their theory of dislocations, basically as is used to this day. Some indirect indications were observed experimentally some years later, but it was not until 1956 onwards that Hirsch was able to use diffraction contrast methods in the electron microscope to confirm the predictions. The use of field ion microscopy finally proved these findings in a scientific hunt almost akin to the search for final confirmation of the elusive Higgs Boson, some fifty years after it was postulated.

As seen in the photos, there are only two fundamental types of dislocation: the *line dislocation* and *screw dislocation*. These are, as their name precisely illustrates, an upsetting of the regular crystallographic structure – producing a line of imperfection that is able to 'glide' through the crystal along preferred planes of atomic close packing. Thus, in fcc crystals the diagonal $[111]$ planes of close packing present routes along which dislocation lines can, on an incremental basis, easily glide. This achieves a net movement through each grain, as the millions of dislocation lines shuffle, atom by atom, across or in a screw-like direction, up the crystallographic structure. These models, at a stroke, explained the observed easy

movement in deformation of metals, compared to the far greater theoretical yield strength based on atomic bond strength.



Photo 102 Model of perfect crystal array, with no defects.



Photo 103 Model of an introduced 'dislocation' (crystal imperfection), as an extra plane of atoms.

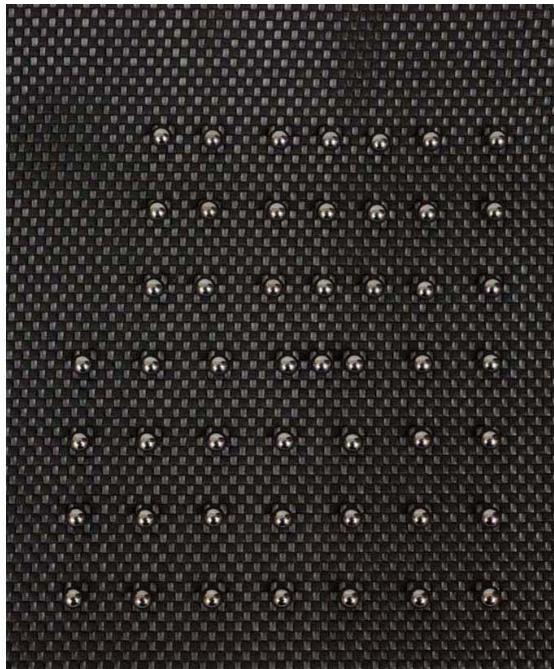


Photo 104 The 'line dislocation' moves through the crystal array.



Photo 105 The 'line dislocation' continues through the structure.

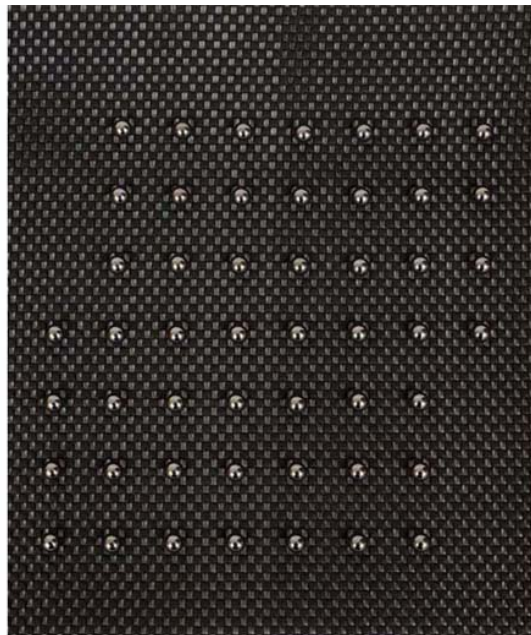


Photo 106 The line dislocation completes its passage, having shifted the fault one 'burgers vector', from left to right.

Line dislocations operate as a moving plane across the three-dimensional crystal, but can be simplified in two dimensions, as

shown above. The movement and direction of the dislocation is defined as the 'Burgers Vector', b , symbol ' $\perp$ ', seen here as a distorting line to the 'normal' crystal. It can move easily along preferred planes, of $[111]$ orientation in fcc structures, a step at a time as it passes from one side of a grain to the other, in much the same way as a bubble trapped during wallpapering can be squeezed easily along the wall until it emerges at the edge of the paper and vanishes (only to reappear elsewhere, according to Murphy's Law!). This is repeated over the great number of dislocations always present in real metals, so that dealing with dislocation movement by their interception at grain boundaries, or 'pinning' by second phase particles in precipitation hardening, is vital as a method for increasing yield strength.

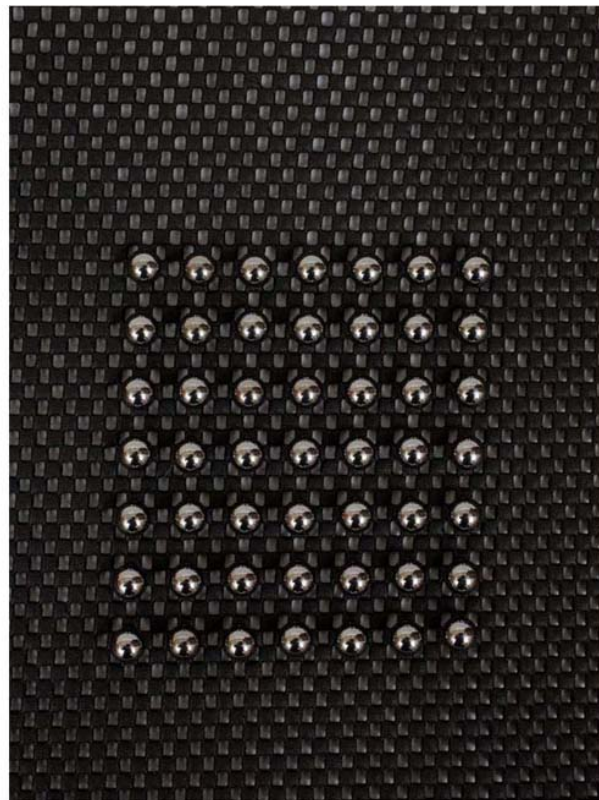


Photo 107 Model of perfect crystal structure, prior to introduction of a 'screw dislocation'.

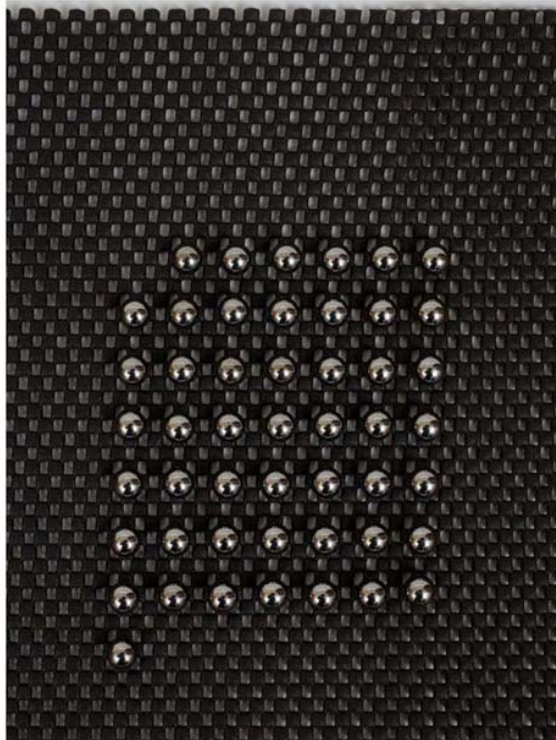


Photo 108 Model of initiation of screw dislocation in a crystal.

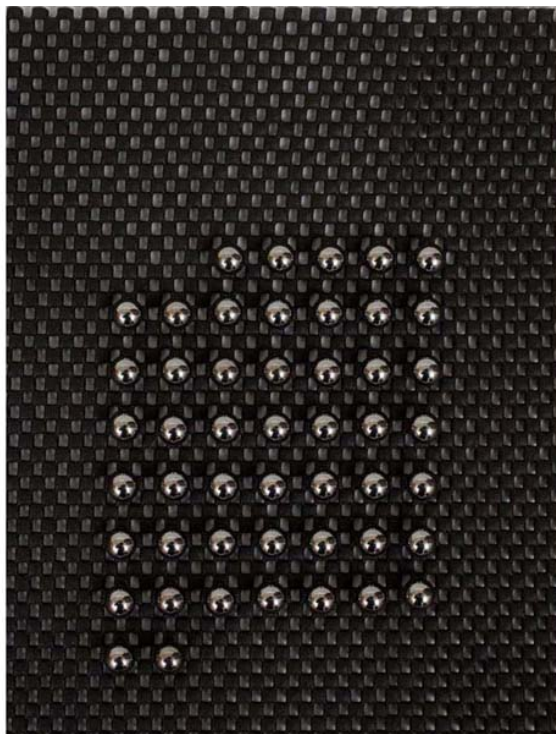


Photo 109 Passage of screw dislocation, vertically through the structure.

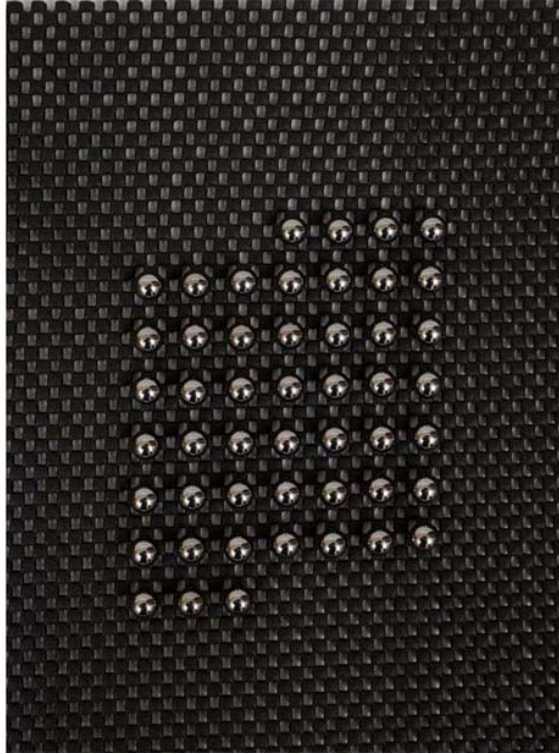


Photo 110 Screw dislocations continues from Photo 109.

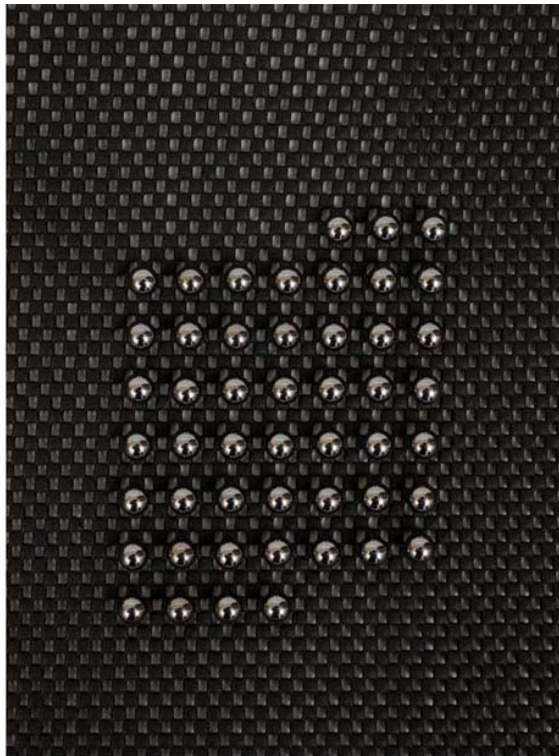


Photo 111 Screw dislocations continues from Photo 110.



Photo 112 Screw dislocations continues from Photo 111.

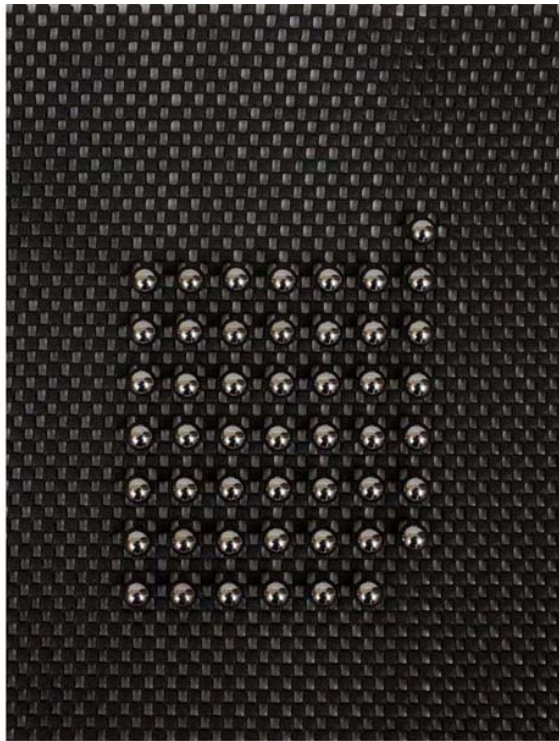


Photo 113 Screw dislocations continues from Photo 112.

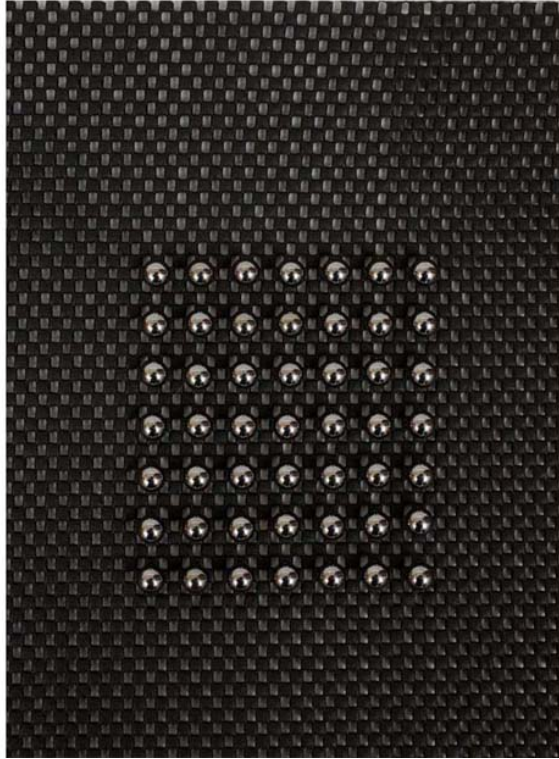
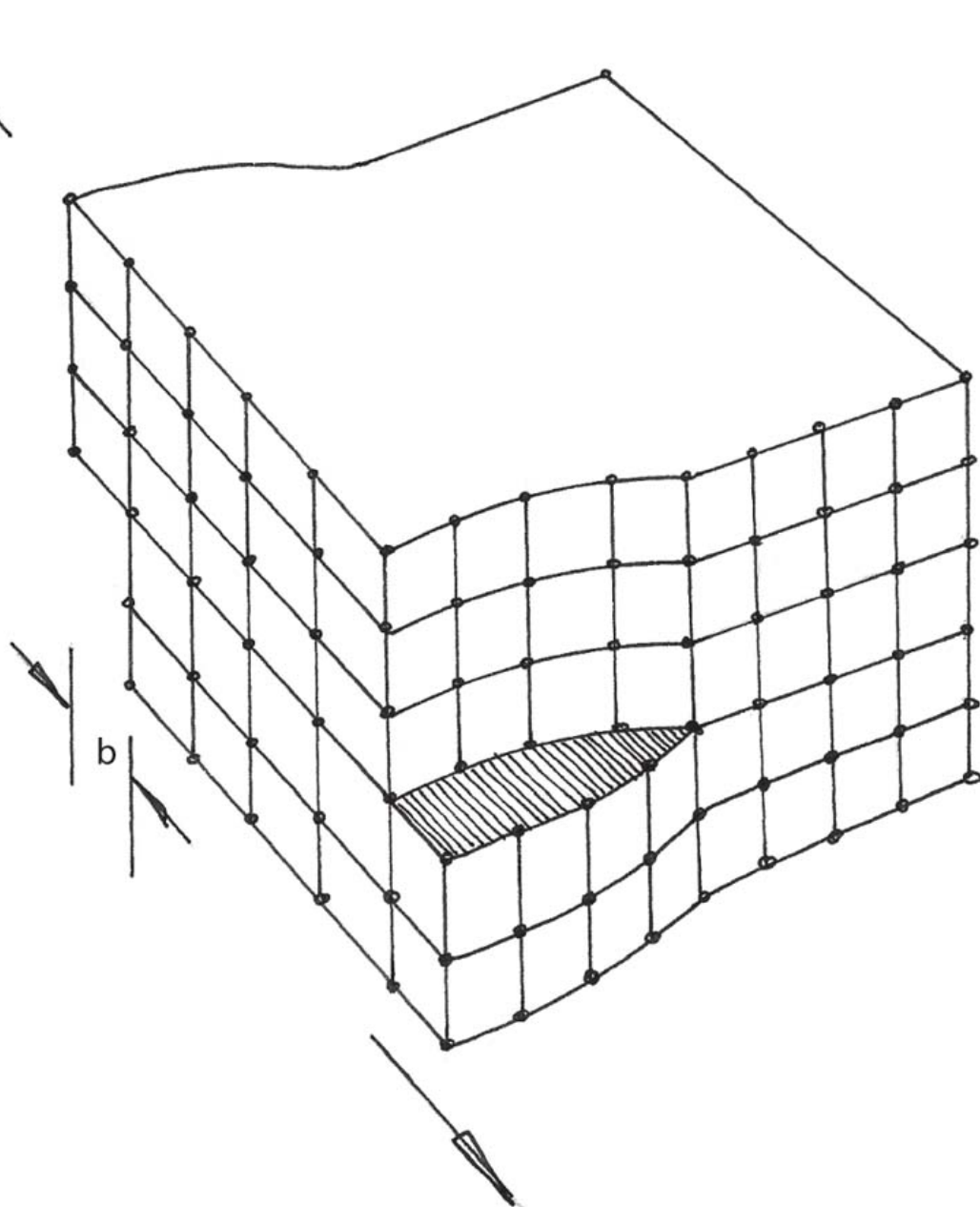


Photo 114 Completion of screw dislocation to top of structure.

displacement



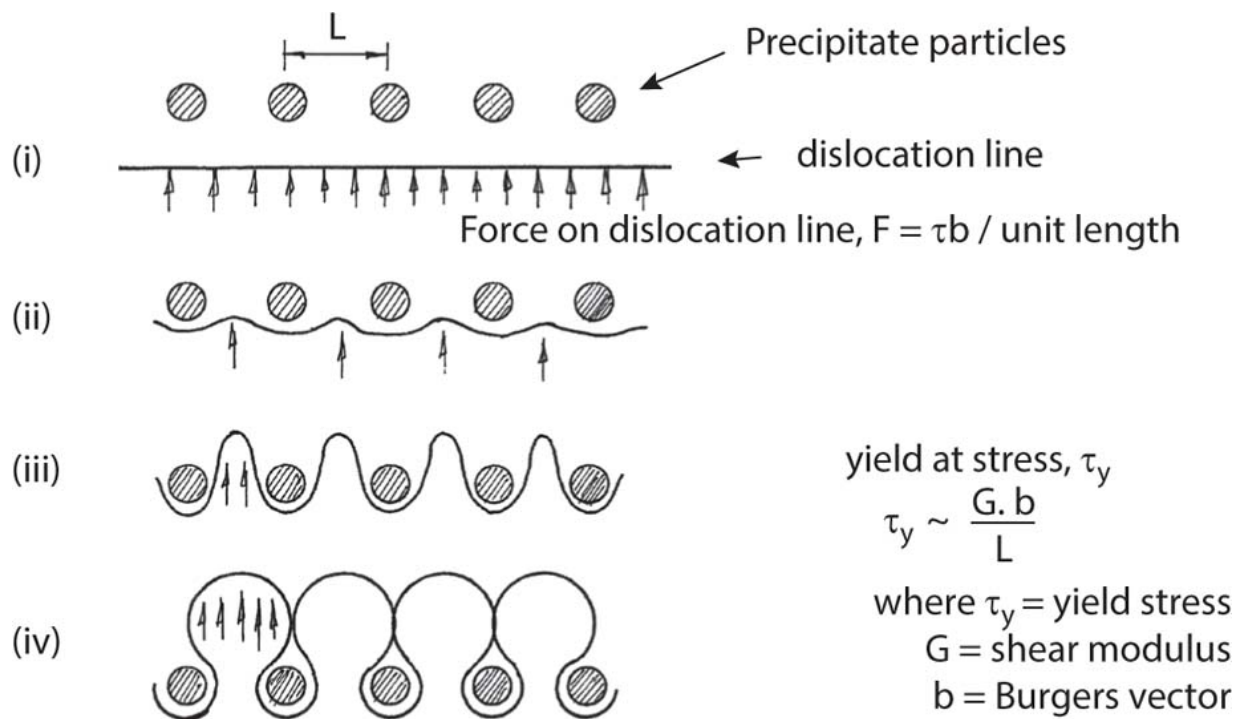
how a dislocation can move swiftly through a grain comprising many crystals, in a 'screw-like' motion

Figure 53 Screw dislocation motion.

Screw dislocations move up through a grain as the crystallographic array accommodates this stepwise helical progression. This is less easy to visualize in two dimensions,

perhaps best seen by the 'plank' analogy where a series of flat planks are moved, one-by-one across a floor, to simulate the glide of the Burgers Vector from one side to the other. The best visualization is from the three-dimensional diagram below.

The presence of a dislocation brings with it a stress field, which can be used to establish the resulting driving force and the effect on the structure. As noted earlier, whilst high strength aluminium was discovered in the early twentieth century, with 'Dural' being produced by adding around 4 per cent of copper to aluminium, it was not until the latter half of the twentieth century that a theoretical understanding was proved from observations of the pinning of dislocations by the sub-microscopic 'clouds' of precipitate (as described later). These ultra-fine particles can be visualized as providing a mesh which interrupts the ever-increasing size of a bubble-gum balloon, causing it to be pinned as it tries to form into numerous small bubbles, effectively resisting its progress, as illustrated below.



- (i) Dislocation approaches particle obstruction
- (ii) Approaching maximum restraint of dislocation
- (iii) At critical (maximum) restraint of dislocation

- (iv) Bowing condition, escape force now below maximum restraint, leading to escape of dislocation line

Figure 54 Precipitation hardening – dislocation obstruction by precipitate particles.

The Frank–Read Source

The theory of dislocations as first postulated still had a fundamental shortfall, requiring a mechanism by which dislocations of sufficient number could be created or regenerated during deformation. Given that it takes a thousand dislocations to generate slip along a line for a minute surface displacement, it would soon be possible to run out of the dislocation supply for the displacements observed. Also, critically, the mechanism of dislocation ‘pile up’ that is fundamental to the increase of strength during strain hardening, would soon consume all of the available dislocations if they were not continually being regenerated. This would lead to cold work *reducing* rather than increasing the dislocation density. The dilemma was resolved when Frank and Read proposed another delightfully elegant model to explain how a dislocation line can bow out under the stress of deformation. This bowing-out continues until it envelopes into a full ring, detaching from the original point sources which remain, to grow another dislocation regeneration cycle, as shown below. Published as a theory by W.T. Read in 1953, the first direct observation of the ‘Frank–Read Source’ came in 1957 to explain how, for instance, cold work increases the dislocation density and strengthening is a consequence of their entanglement and subsequent loss of slip facility. This is readily observed, as shown in working austenitic stainless steel at room temperature.

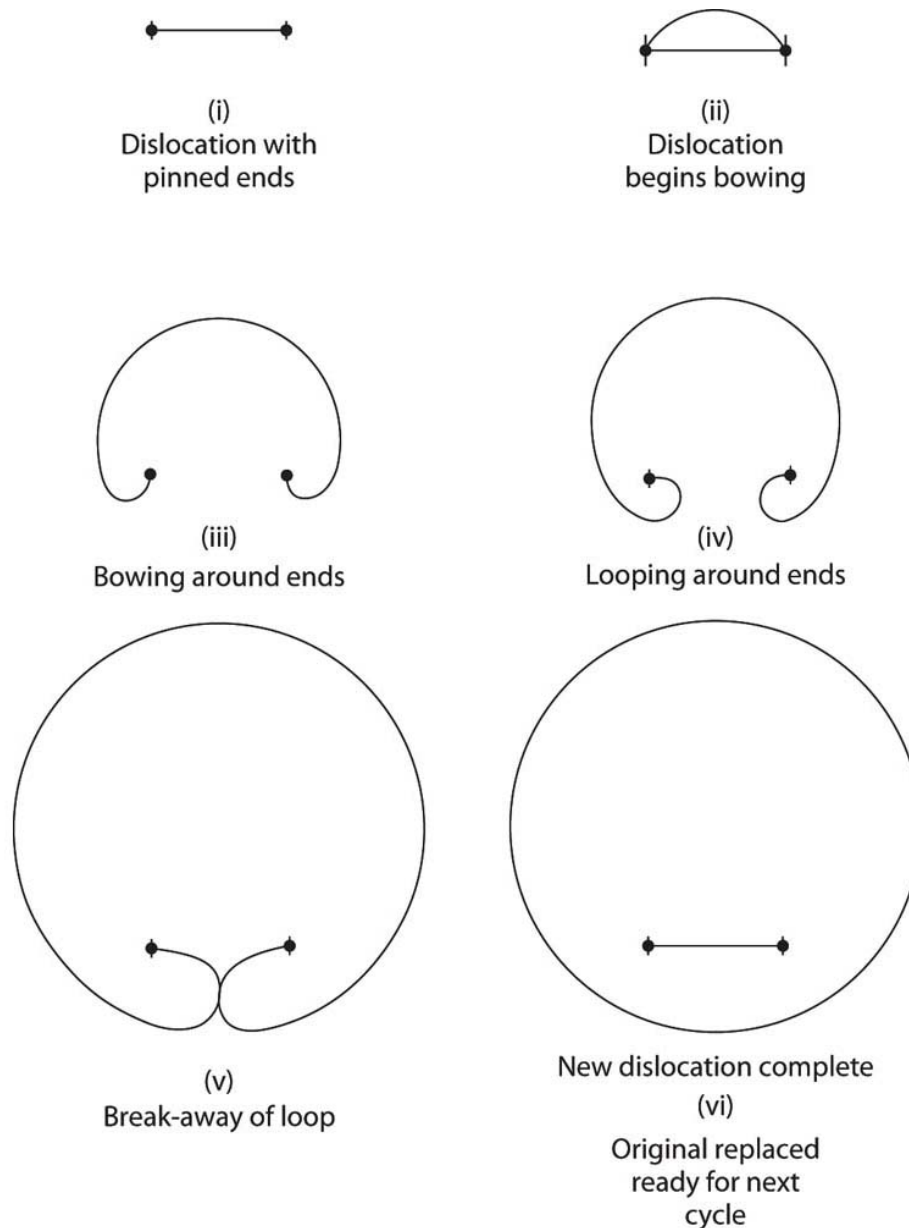


Figure 55 The Frank-Read source for Dislocation Generation.

Solidification

Apart from metals produced by additive manufacturing – or powder-metallurgy where fine metallic powders are compacted and sintered by a number of commercial processes – all metals originate from a casting. As with the weldments mentioned before, this involves transferring liquid metal across an arc plasma, or pouring from the

liquid into a solid form; it is deserving of some attention as the fundamental process of solidifications and crystal growth.

Grain size control in castings

The general solidification process, described earlier in [Chapter 2](#), 'Castings', showed the classic three-stage process, progressing from the initial chilled crystals at the wall of mould; leading to dendritic growth that continues as long, columnar grains into the heart of the melt; with under-cooling and spontaneous nucleation finally producing equiaxed, more regular and spherical, crystals. Grain size variations, as cast, are largely influenced by the variations of cooling rate in the mould.

Such a casting has a reasonably satisfactory structure in the centre but this is not the case for the area of large columnar grains, as the aim should be for a fine-grain casting throughout, to deliver the optimum properties from the benefits of small grain structures, as described earlier. The usual solution is to modify the melt by deliberate intervention, employing 'inoculation'. This is done either just before or during the pouring process, providing a catalyst for the widespread nucleation sites required for the spontaneous growth of small equiaxed grains throughout the melt. This is a highly effective method of overcoming the natural solidification process of a mixed structure, only some of which is satisfactory. It should be noted, however, that as a highly sensitive process the business of inoculation is far easier to describe than achieve in practice, relying on the craft and skill of the foundryman.

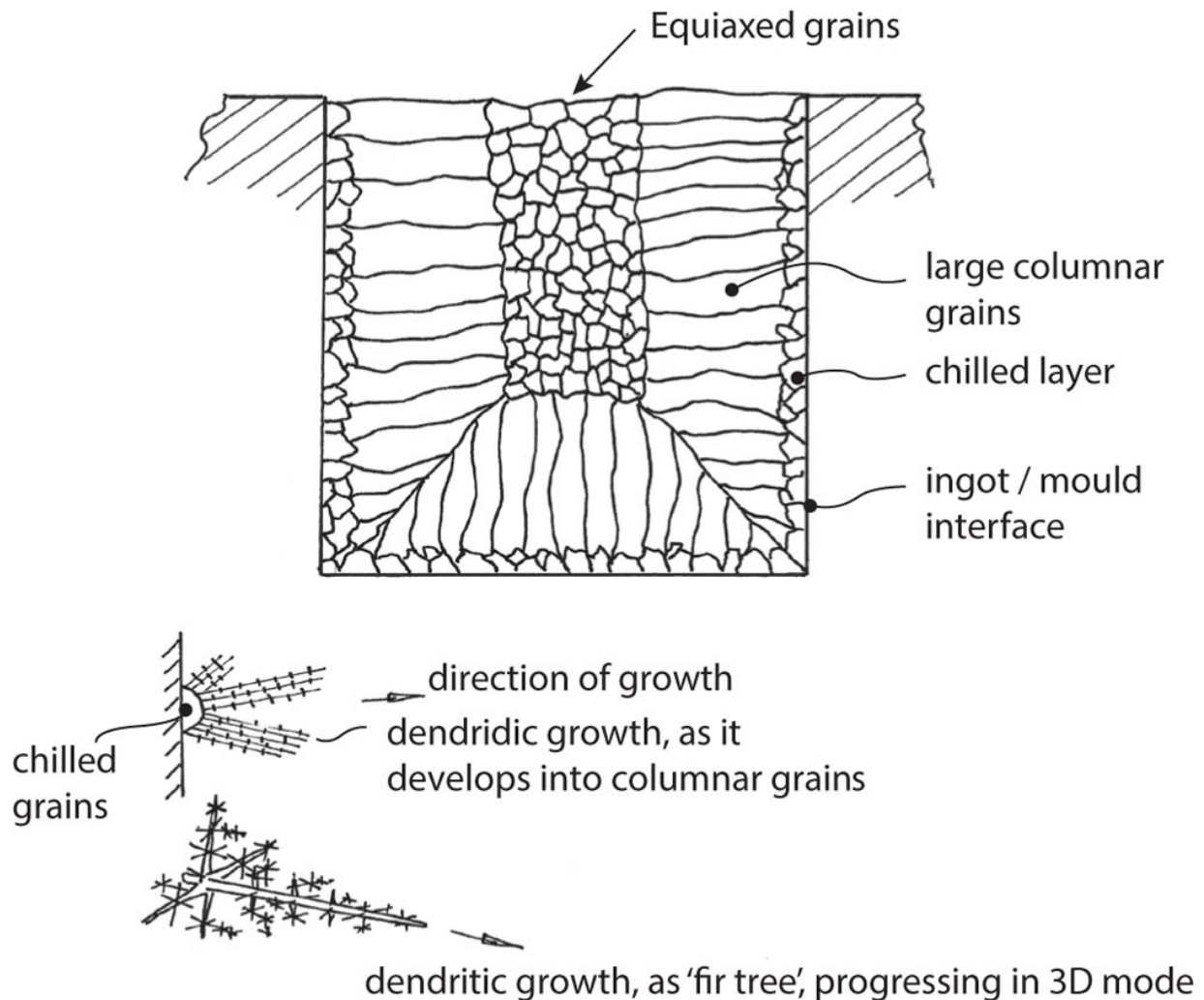


Figure 56 Casting solidification patterns, typical of unmodified structures.

Re-crystallization

Returning to the theory of dislocations, we can now also explain the well-known behaviour of 'recovery' and 're-crystallization'. This enables a metal that is strain-hardened not only to be stress relieved but also, with proper control, to actually be brought back to its original condition, and even be grain refined, by a process of controlled deformation and annealing. The temperature of annealing is $\sim 0.6 \times T_m$, where T_m is the melting point in $^{\circ}\text{K}$ (where $0^{\circ}\text{K} = -273^{\circ}\text{C}$, absolute zero), so a 'sub-critical', or 'process' annealing treatment around $600\text{--}650^{\circ}\text{C}$ (below A_1 , lower critical temperature) is common for steels.

It has been seen earlier that strain hardening, as cold working, produces a highly significant increase in dislocation density, from the annealed, soft, condition. Typical microstructures contain $\sim 10^6$ (a million) dislocations per mm^2 , and cold working can increase this to $\sim 10^9$ (a billion) dislocations per mm^2 . The materials we commonly deal with are often deformed by rolling, some 50 to 80 per cent cold working being possible before their plasticity is exhausted and fractures appear.

The soft and ductile annealed condition, obtained from this slow heating and cooling process, results in the minimum dislocation density. When worked above a critical amount of around 5 per cent deformation, the metal can then be annealed again. At this low level of deformation only the 'recovery' process is achieved. This is a submicroscopic process where the dislocations begin to rearrange themselves to facilitate the growth of low-angle grain boundaries. These are relatively subtle changes and so the mechanical properties of ductility and hardness are little improved, and dislocation density not greatly reduced.

Further deformation, prior to annealing, increases the dislocation density. A rolling operation of 50 per cent reduction has the effect of raising the dislocation density, and with it the strain energy associated with the dislocation formation, to around 2×10^{-3} Joules per mm^3 (2 mJ mm^3). This provides the driving force to take the structure beyond the stage of recovery and into *re-crystallization*. In the latter, this dislocation energy promotes the nucleation of new, equiaxed, grains from within the old, deformed grains. This continues until, almost magically, the cold-worked and deformed structure is replaced entirely by a new, fine-grained structure. This structure is obtained from annealing, after final cold working. The dislocation density is restored to the original level, with the internal energy being given up to the re-crystallization process, as shown below.

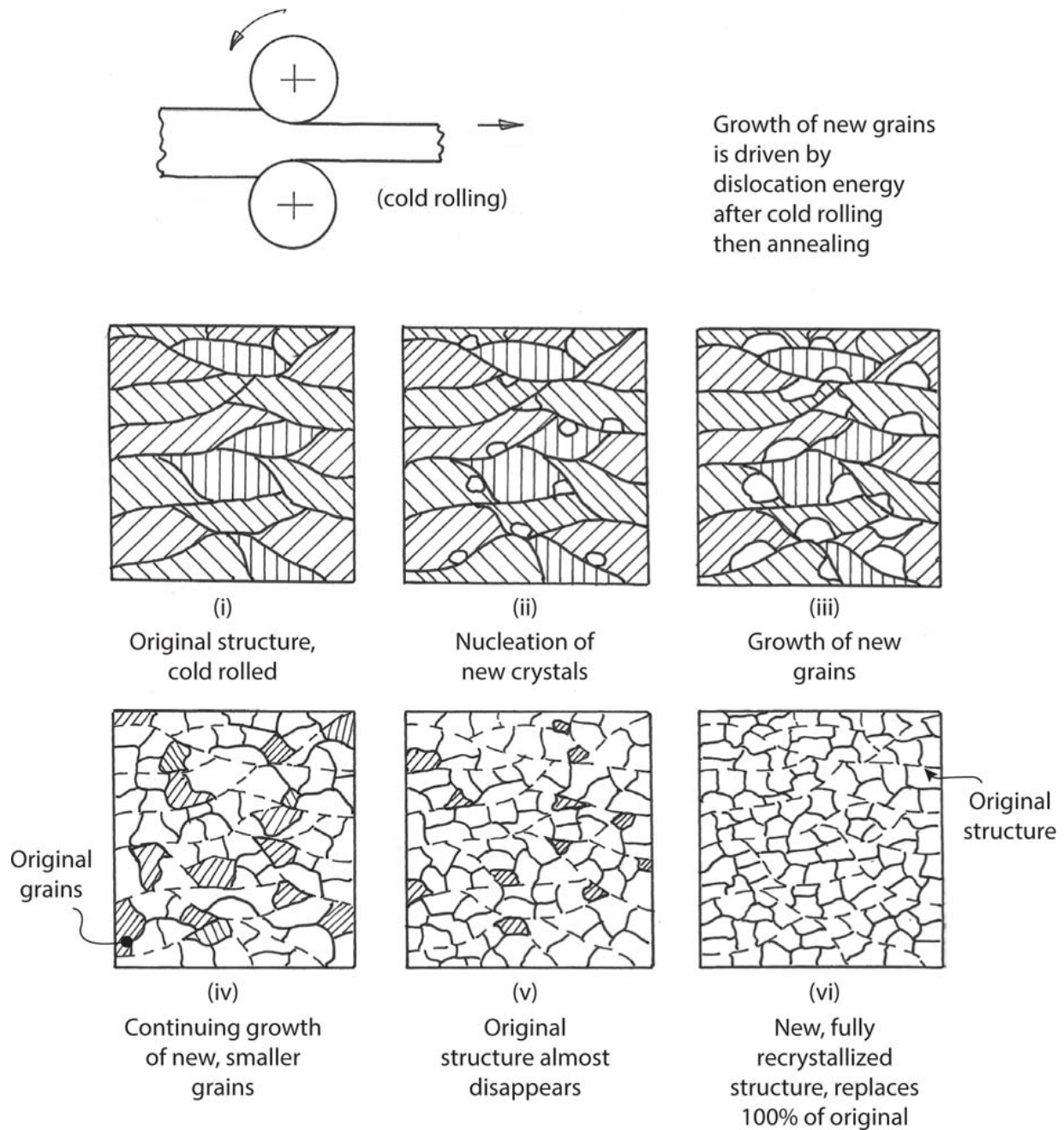


Figure 57 The process of recrystallization.

The use of controlled working, followed by annealing, can therefore achieve not only the expected stress relieving of the strained material, but also restores the microstructure to the optimum fine grain condition. This treatment is utilized particularly in rolling operations, and can be employed in the final stage as the material is reduced from a slab to thin sheet. Note that, were a large

and elongated grain structure to be produced in the final thin sheet, this would also cause great problems in effecting reliable forming. As seen earlier, the basic facility of metals to behave 'anisotropically', unaffected by the direction of working, would be jeopardized. If grains were large, and single grains existed through the sheet thickness, this could result in an irregular response to shaping, with ripples across a car body panel, or refrigerator door.

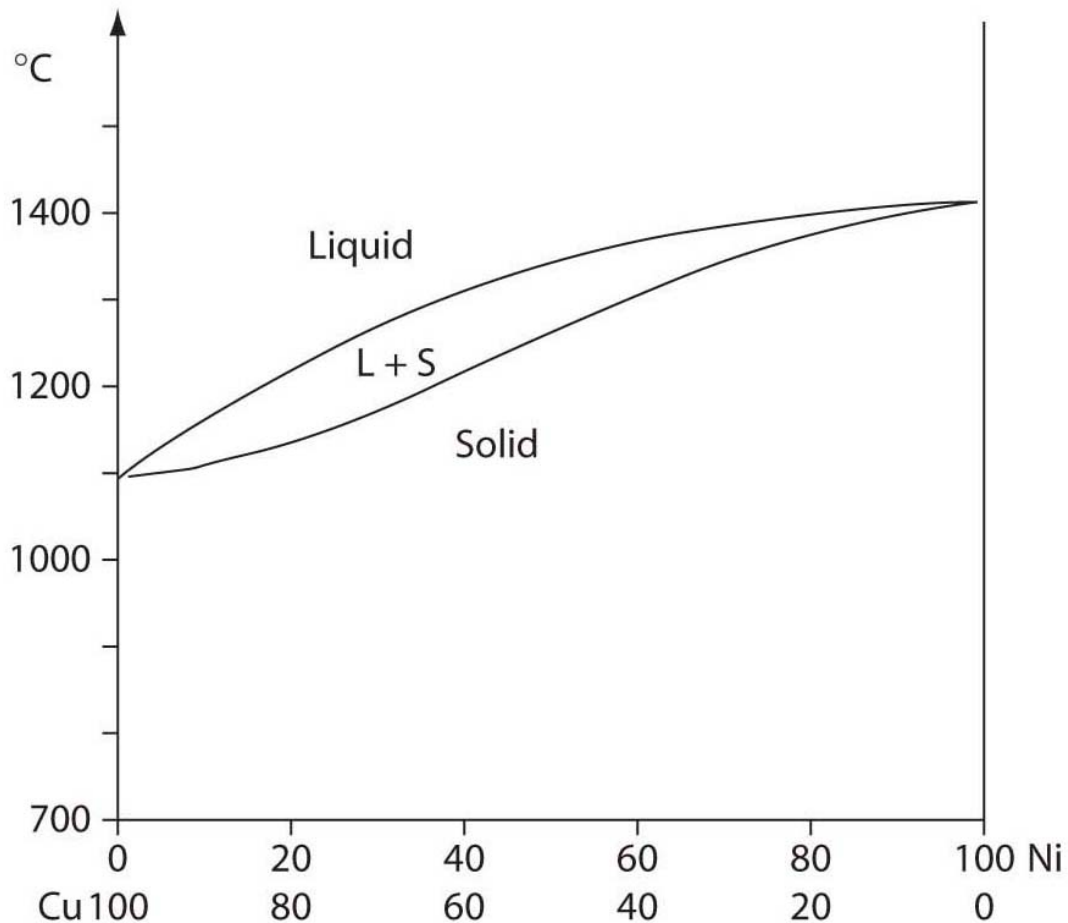


Figure 58 Cu-Ni equilibrium diagram.

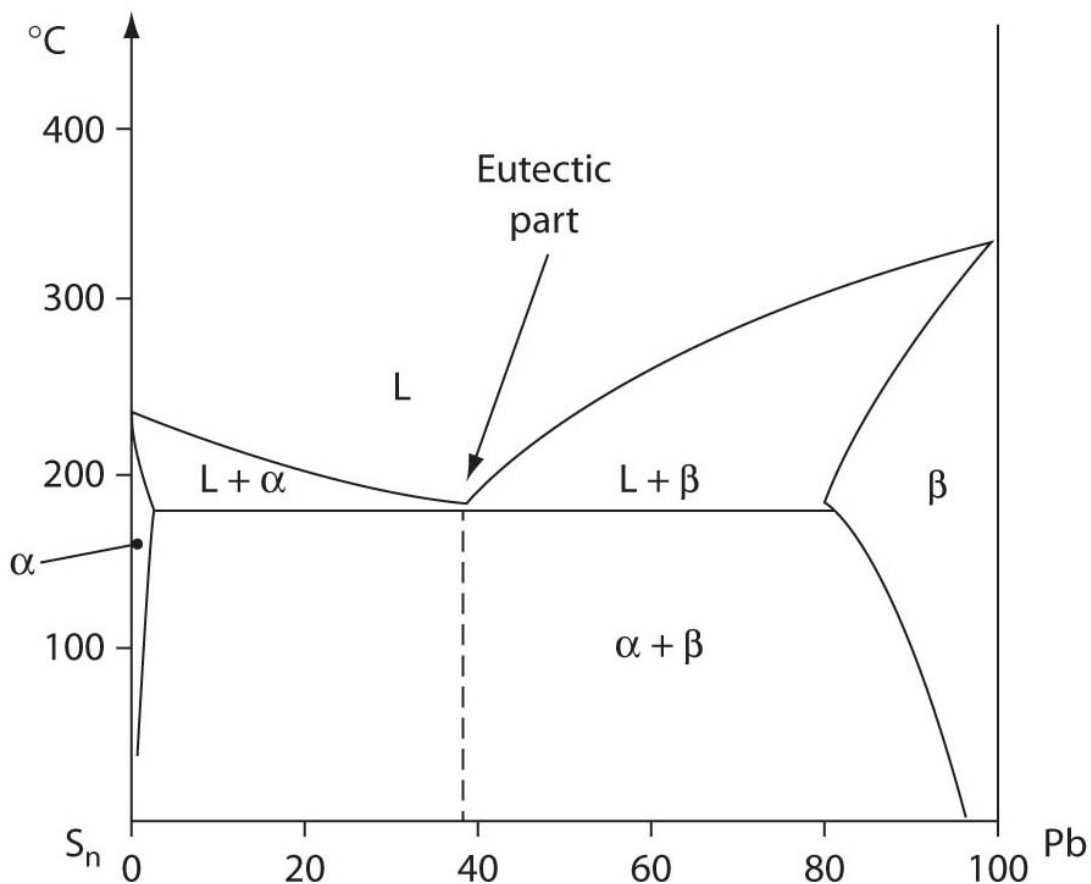


Figure 59 Sn–Pb equilibrium diagram.

EQUILIBRIUM DIAGRAMS

These are a vital tool for the understanding of the mixture of pure metals, alloys, which constitute engineering materials. Although the pure metals have significant applications where corrosion, ductility, electrical and thermal conductivity requirements predominate, as noted previously in [Chapter 2](#), they fail to provide the mechanical properties of strength and toughness demanded by common engineering applications.

It is therefore vital to be able to consider the effects of alloying, from the most simple two-component system where each element is completely soluble in the other ([Figure 58](#)), to the case where the two components are fully soluble in the liquid but only partially

soluble in the solid state ([Figure 59](#)). The third fundamental type of equilibrium diagram is the 'peritectic reaction' ([Figure 60](#)), and finally the important systems containing intermediate phases, known as 'intermetallic compounds', all shown below.

These equilibrium diagrams serve to illustrate the principle, but care is required in practical applications, for a couple of important reasons. Firstly, equilibrium diagrams are, explicitly, only applicable where complete equilibrium has been achieved in the system. In practice this may take a very long time, or may effectively never be fully realized. Secondly, the examples deal with only two-component systems, and in practice alloys are comprised of more, often many more, elements. Whilst a three-component system, ternary diagram, may be drawn reasonably easily, as in [Figure 64](#), the third dimension can only practically be extended over a short range of the third element. Likewise for the further dimensions applicable with each additional element, as it becomes impractical to consider other than a single composition.

Notwithstanding these limitations, with care and experience the two-component equilibrium diagram, perhaps redrawn at a single value of any further elements, can provide an invaluable guide into selection of practical heat treatment holding temperatures, and the consequence of rapid or slow cooling, as the phases shown on the equilibrium diagram are followed or traversed. These diagrams are critically important in identifying the composition where rapid cooling can arrest a high-temperature phase, retaining it in the metastable (frozen) state at room temperature, after which controlled heating can provide 'precipitation hardening', or 'tempering' of martensite, as seen later.

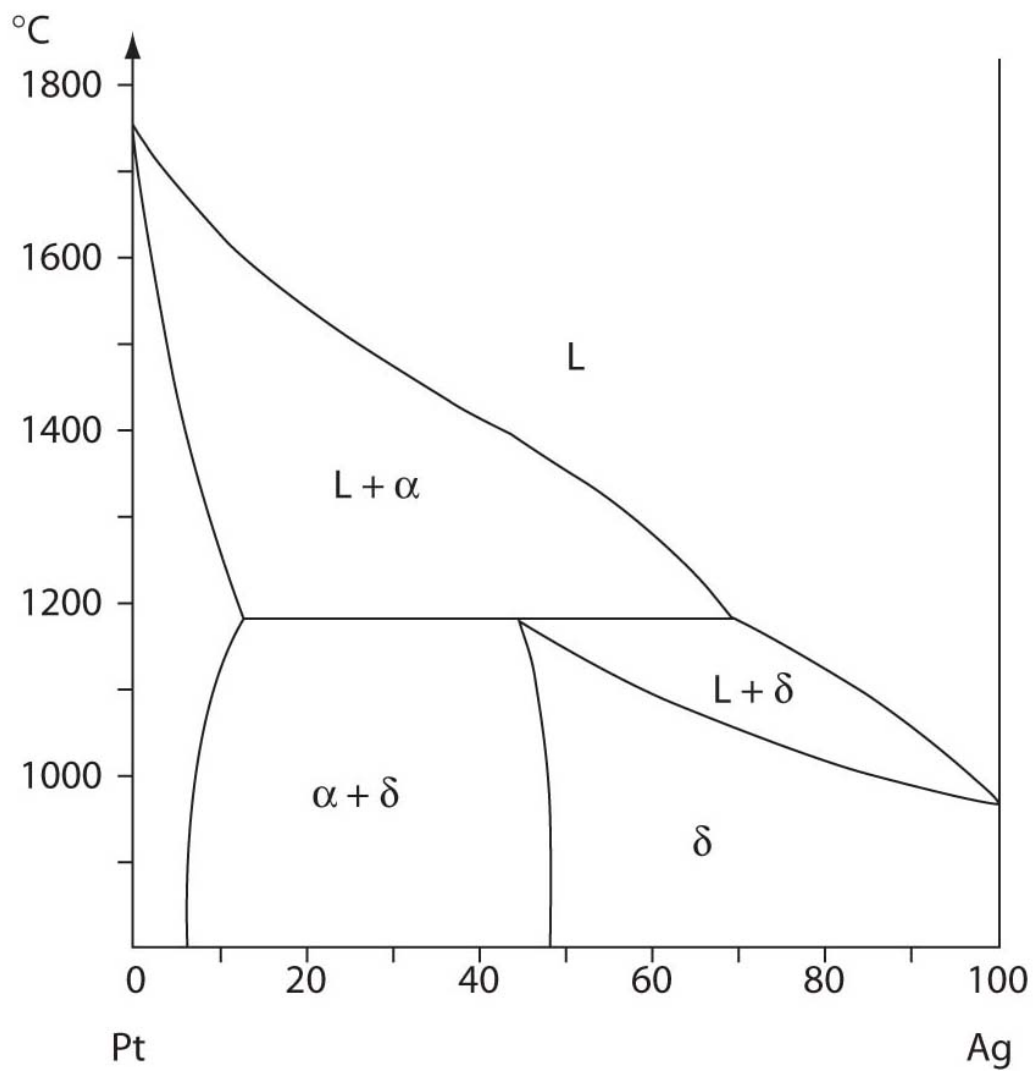


Figure 60 Pt–Ag equilibrium diagram.

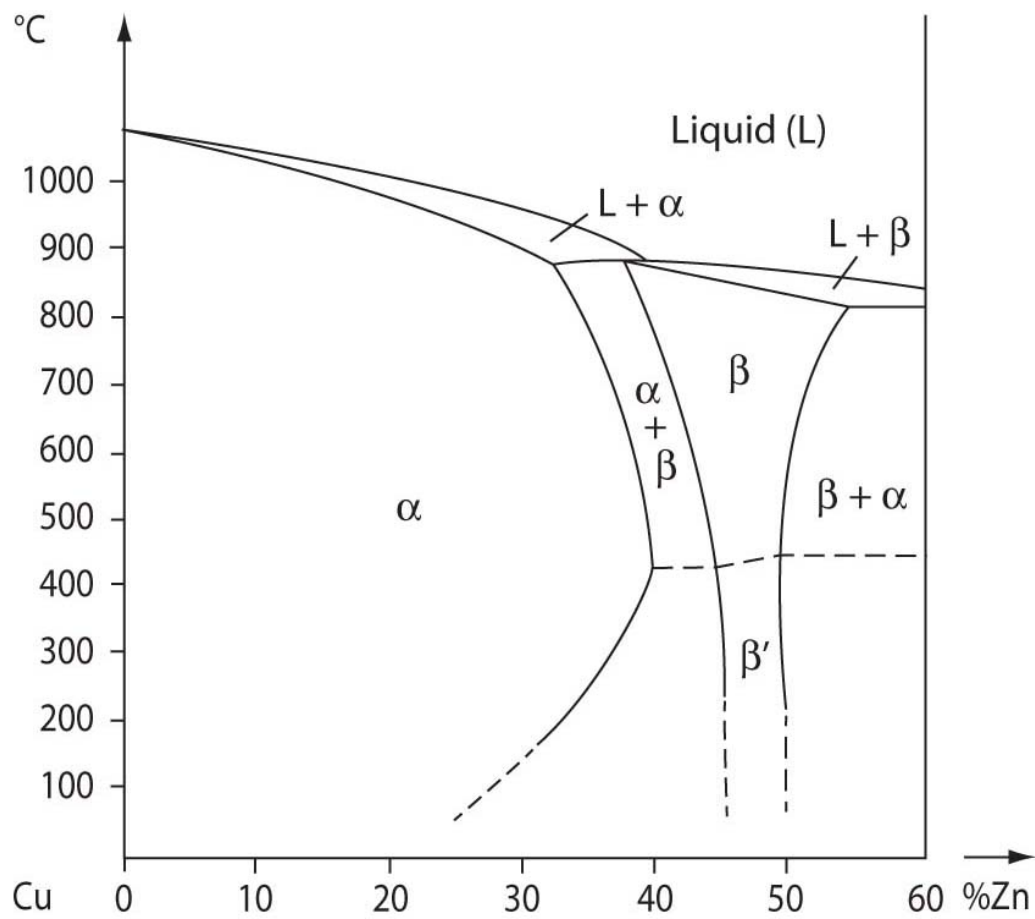


Figure 61 Cu–Zn equilibrium diagram.

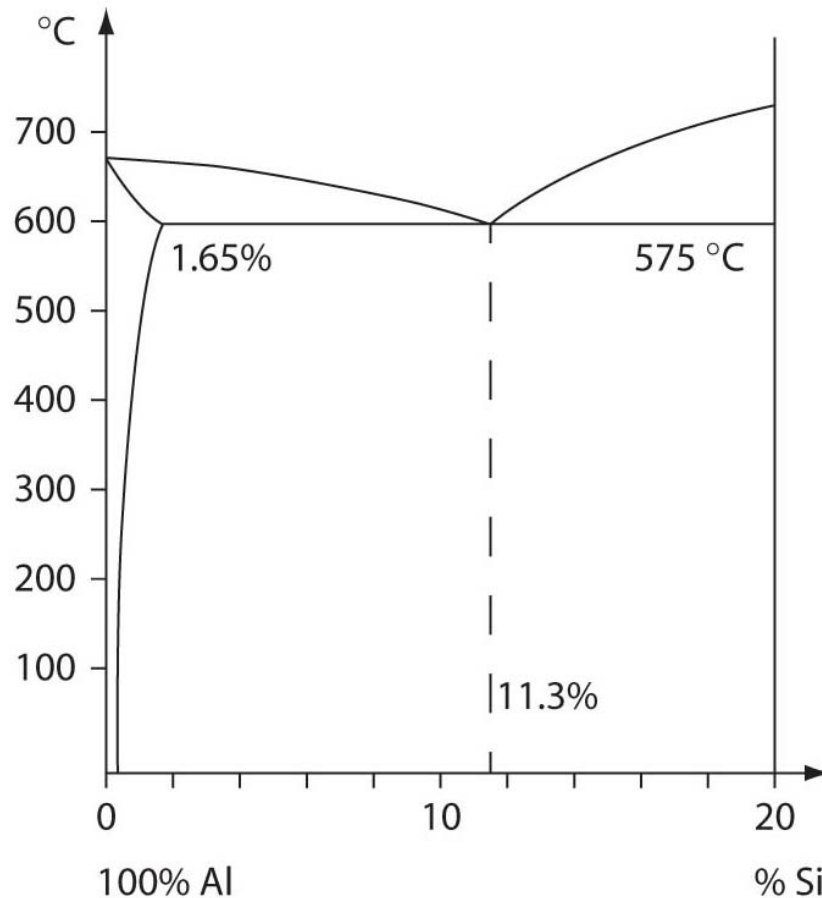


Figure 62 Al-Si equilibrium diagram – for cast alloys.

Often, in practical applications, the limitations of the two-component system can be virtually neglected, for instance in common solder where lead replacement elements still provide a similar diagram that makes sense, as above. The important features include how the elements combine to form a single freezing point at the 'eutectic' (derived from the Greek 'eutektikos' – able to be melted easily). This mimics the freezing behaviour of a pure element which occurs at a single temperature, whereas the two-element freezing takes place over a range of temperatures between liquid and solid. This is the 'pasty' range as seen in the equilibrium diagram, where solid and liquid co-exist. This is observed in (non-eutectic) solder, which is seen to begin and finish solidification over a range of temperatures (see Figure 59).

Surely the most important alloy system, that of steel and iron, is well described by the FeC equilibrium diagram. It is of great practical use, despite the addition of further elements, because of the overwhelming impact of carbon on the alloy's behaviour, as seen later.

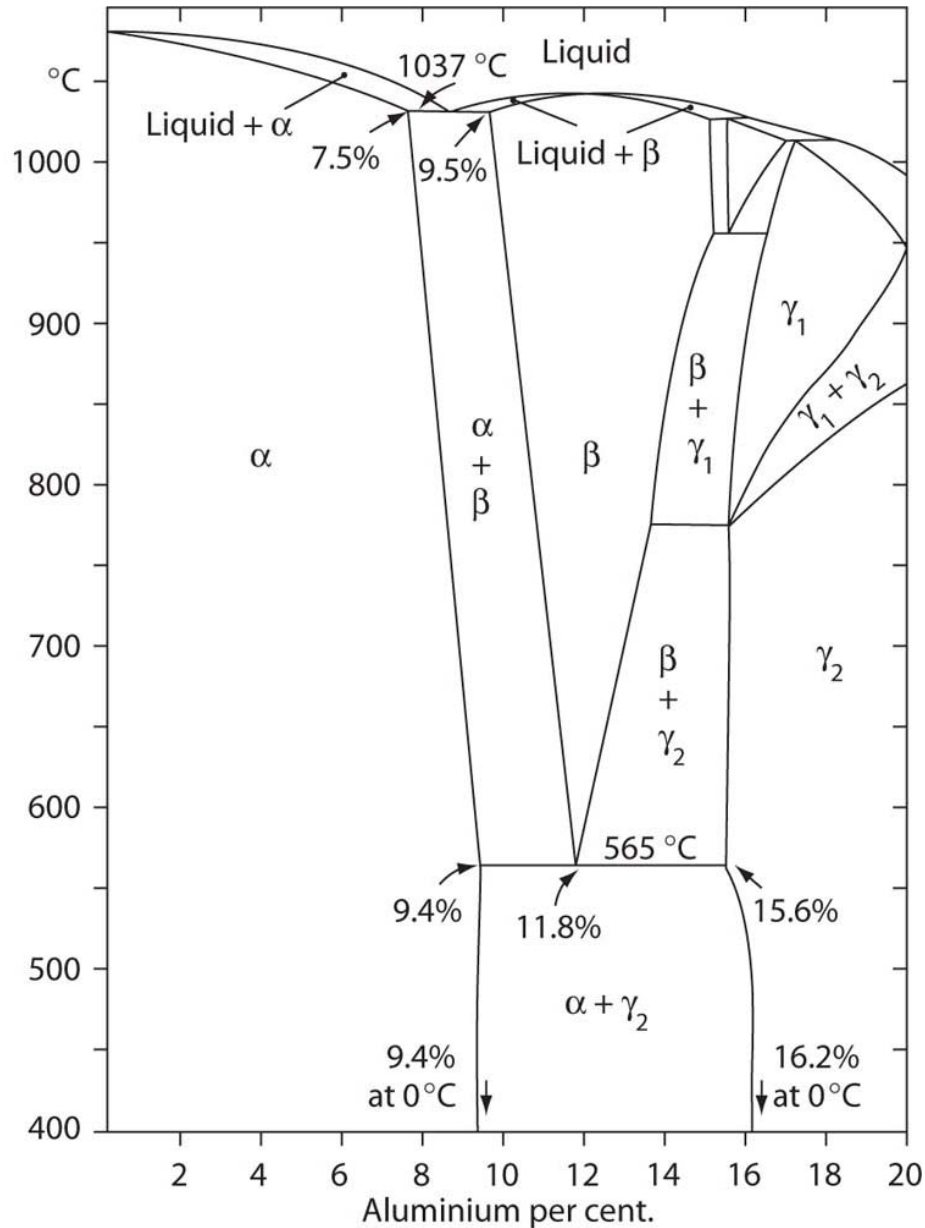


Figure 63 Cu-Al equilibrium diagram (aluminium bronze).

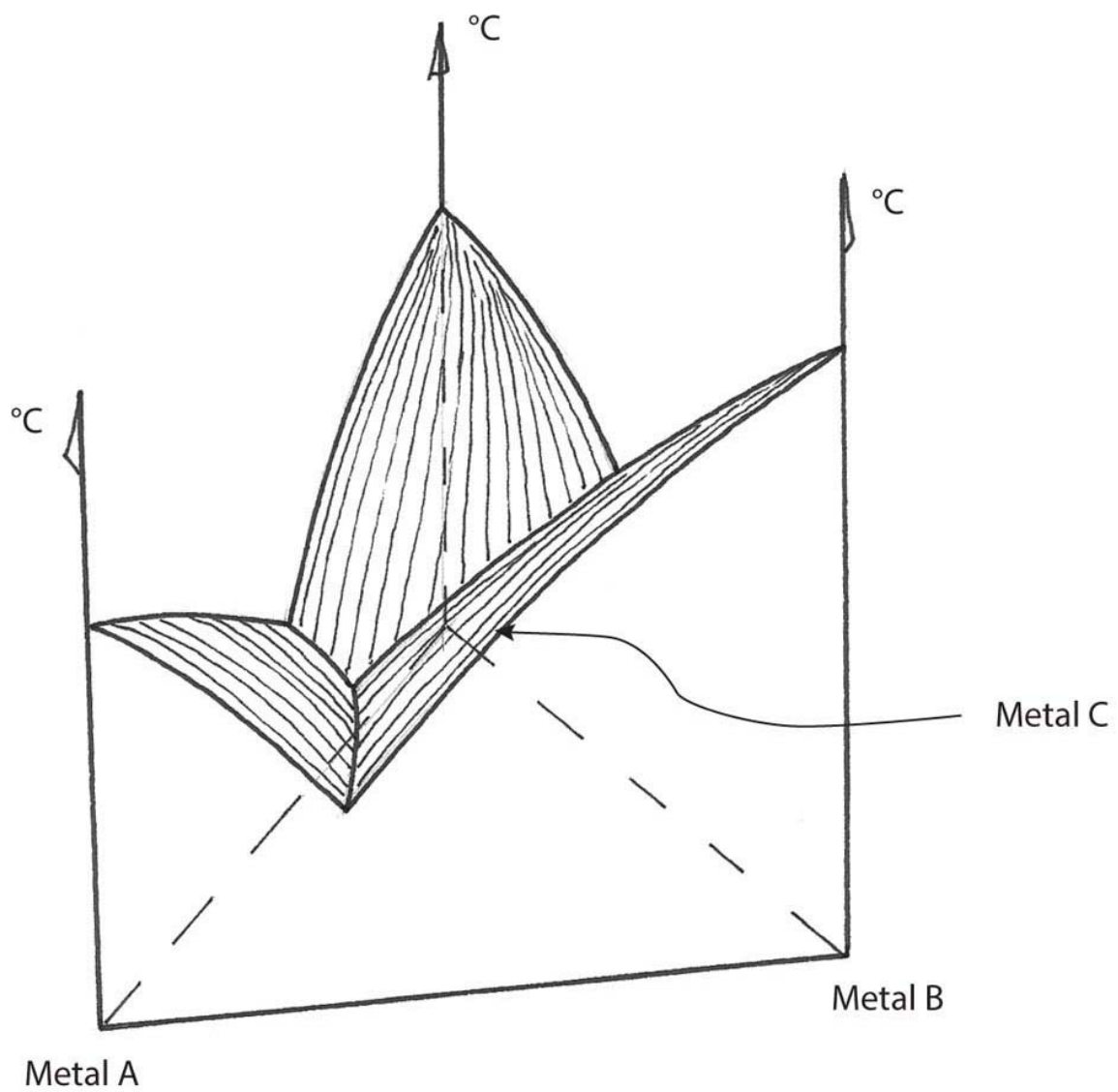


Figure 64 Ternary diagram (equilibrium diagram in three dimensions).

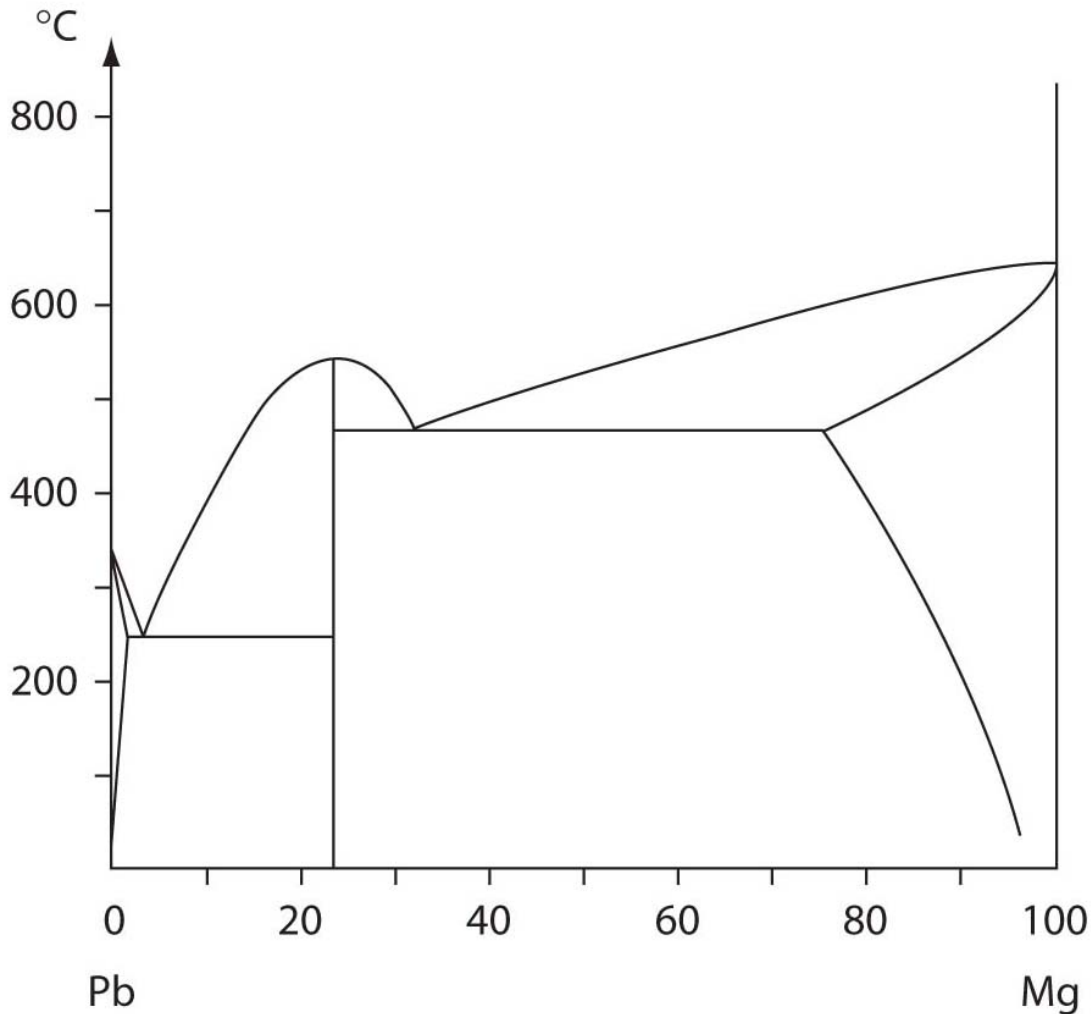


Figure 65 Pb–Mg equilibrium diagram.

In many cases, the equilibrium diagram is a satisfactory guide in predicting the microstructure of an alloy. From a given composition, when slowly cooled to obtain near-equilibrium conditions, growth of the phases dictated by the solidification process can be predicted from the diagram. Where a eutectic point exists, a composition coinciding with this will form a complete and intimate, homogeneous, mixture of the two elements (as shown in [Figure 59](#)).

In the peritectic reaction, the already solidified phase reacts with the remaining liquid to produce an entirely new phase, derived from its Greek origin, 'peri' (around), where the solidified phase is surrounded by the reaction product (see [Figure 60](#)).

It is frequently found that two elements form both limited solid solutions, and combine at particular compositions to form intermetallic compounds. This is illustrated in [Figure 65](#), where a single intermetallic compound is formed at ~24 per cent Mg. This complicates the equilibrium diagram and has effectively rendered it as two abutted versions of that shown earlier with a eutectic ([Figure 59](#)).

STRENGTHENING AND JOINING THE LIGHT ALLOYS

Whilst there are some fourteen pure light metals, apart from their use as alloying elements or in very specialized applications, only three are of concern as common engineering materials: aluminium, titanium, and magnesium. Their general characteristics of density, strength and stiffness have already been discussed, in [Chapter 2](#), but here we are concerned with the strengthening mechanisms that have brought them to prominence, to such an extent that one of their number, aluminium alloys, find very extensive application in many and diverse products, ranging from the drinks cans to structural elements that challenge the traditional use of steel and cast iron in vehicles. The virtue of their corrosion resistance can be matched to a strength-to-weight ratio comparable to low-alloy steels, based on the strengthening mechanisms described next. One of the few down-sides of the use of such common light alloys is the significantly lower stiffness derived from their inferior E (modulus of elasticity) value.

With their increased usage, the aluminums have become more readily available, and economically viable for the home worker – either for reclaimed, or possibly, new material. This is unfortunately not yet the case for titanium. With an enviable corrosion resistance, titanium has applications as an inert biomaterial; it is used in the petro-chemical industry; and even in the front stages of turbine blades for jet engines. Sadly for the amateur engineer, these are all specialized fields, and relatively insensitive to material cost. Similarly, magnesium, which has the lowest density of the three and, when

strengthened, has the potential for the highest strength-to-weight ratio, is less widely used and remains expensive. It is also not without fabrication problems, such as difficulty in welding without degrading its strength.

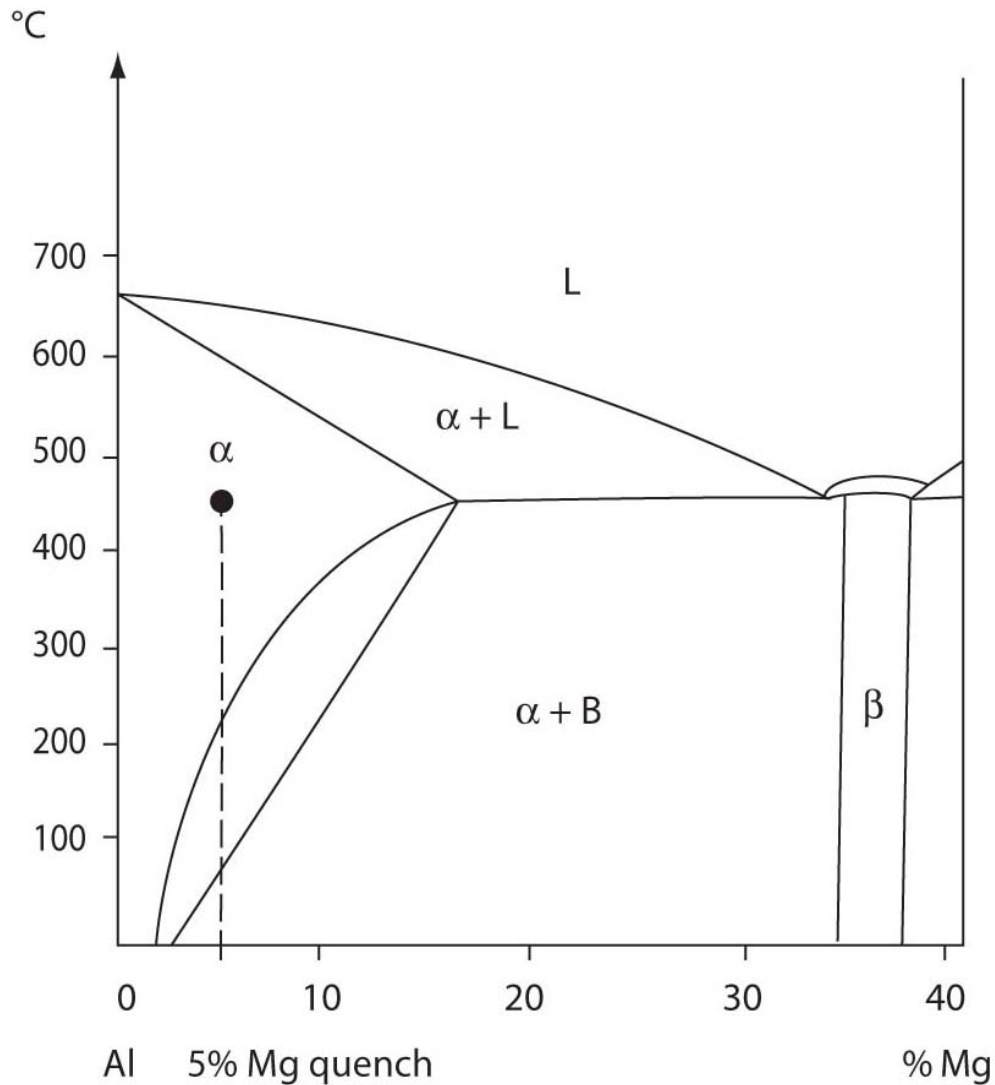


Figure 66 Al-Mg equilibrium diagram.

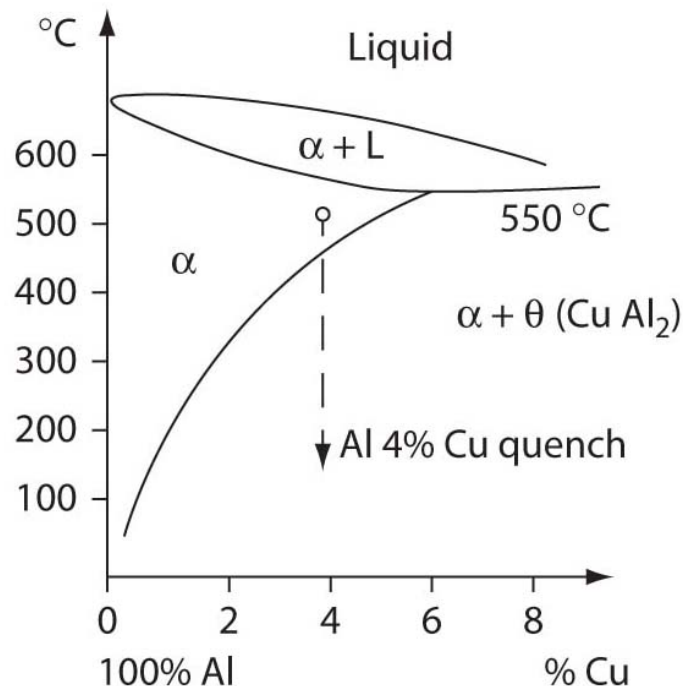


Figure 67 Al–Cu equilibrium diagram – eg. Dural.

Methods of strengthening

All of these systems, Al, Ti, and Mg, can be strengthened by the three basic approaches: solid solution hardening; precipitation hardening; and work hardening. The same principles apply across the three systems, so a more detailed review of just aluminium follows.

Solid solution hardening

This is a mechanism available to all alloyed aluminums of the 1000- to 7000- series, as also seen in [Chapter 2](#). It is of particular benefit in the Al-Mg, 5000- series, non heat-treatable alloys. The potential for solid solution hardening derives from the Al-Mg equilibrium diagram, shown in [Figure 66](#), with its limit of <2 per cent Mg in Al at room temperature. An Al-5 per cent Mg alloy, heated to 400–500°C and quenched, provides a metastable, super-saturated microstructure. The resulting microstructural mismatch and internal

strain makes dislocation movement more difficult, hence increasing yield strength, as per [table 34](#). This shows the effect upon strength of the supersaturation of Mg in Al, and means that alloys such as 5083 aluminium can be fusion welded without significant loss of strength in the adjacent material, the HAZ ('heat affected zone'), which is a weakness that afflicts the heat-treated alloys, such as the weldable 6000-series when left in the 'as-welded' condition, as noted later (see 'Joining' later in this chapter).

The popular, high strength Ti alloy, Ti 6Al-4V, and Mg alloys with 2–5 per cent Al or Zn, are also solid solution hardening alloys. Whilst the other aluminiums benefit from this mechanism, they may also utilize precipitation hardening, a mechanism so potent that it swamps that derived from solid solution hardening.

Table 34 5000-series aluminium alloys (Al-Mg solid solution hardening).

Alloy	Mg (%)	Yield strength (0.2% PS N/mm2)
5005	0.8	40
5050	1.5	54
5052	2.5	90
5083	4.5	142
5456	5.1	150

Precipitation hardening

Also known as 'age hardening', this mechanism was first discovered, as noted earlier, in the 'duralumins' over a hundred years ago. These form the 2000- series of the Al-4 per cent Cu family of alloys, an equilibrium diagram of which is shown in [Figure 67](#). The extremely small solubility of Cu in Al at room temperature, compared to ~7 per cent at 550°C, provides the opportunity for quenching a 4 per cent

Cu alloy from $\sim 550^{\circ}\text{C}$ to produce, once again, a supersaturated solution. However, this is just the beginning of a complex, sub-microscopic sequence that was discovered long before it was scientifically unravelled in the mid-twentieth century, with the aid of dislocation theory, and the electron microscopy able to probe the nanometre world of the fine intermetallic precipitates.

The practical process of quenching from the fully α -phase to room temperature, and the 'ageing' by low temperature heat treatment, around 150°C for ~ 100 hours, belies the complexity of the processes involved. A four-stage process, including an 'over-ageing' which would eventually negate the spectacular strength gains from the beginning, is shown in [Figure 66](#).

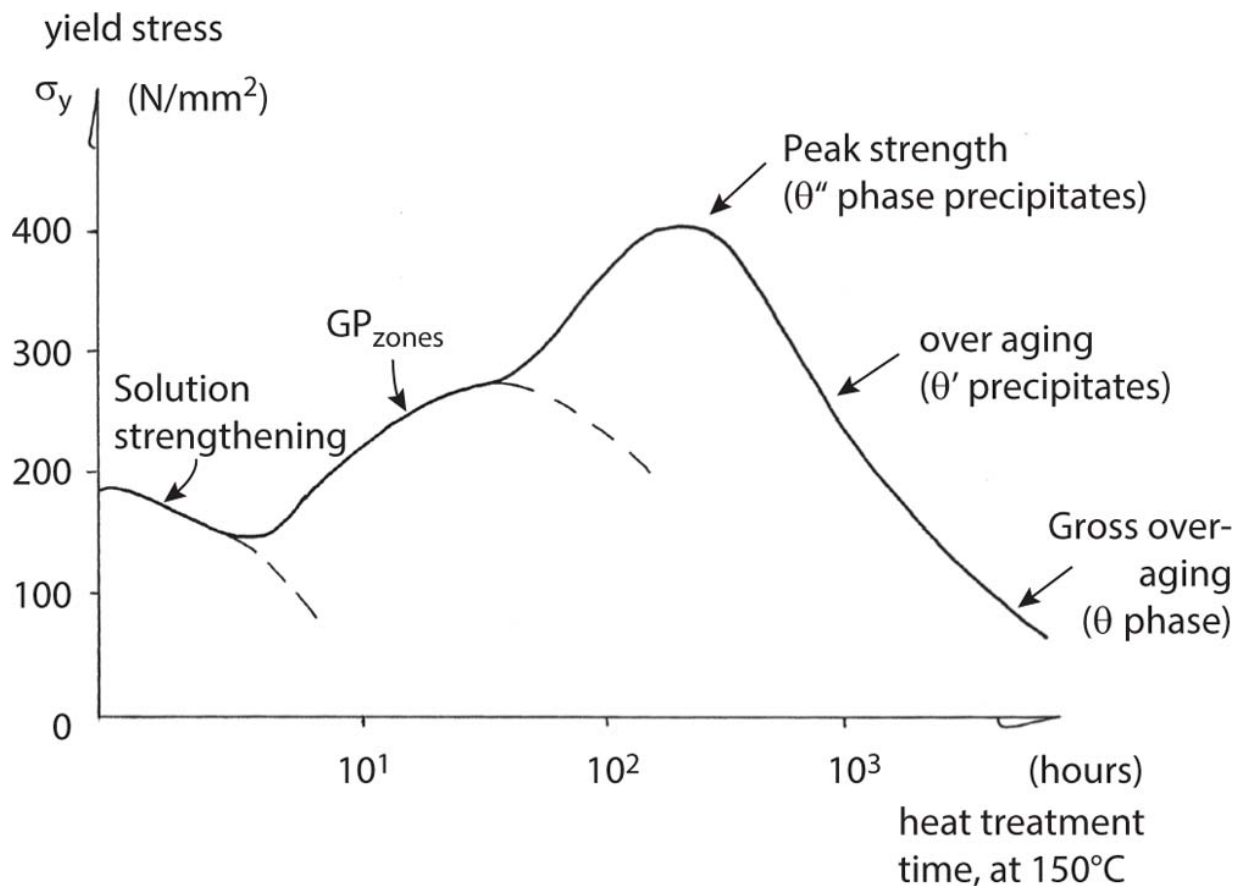


Figure 68 Precipitation hardening for 2000-series Cu-Al alloys.

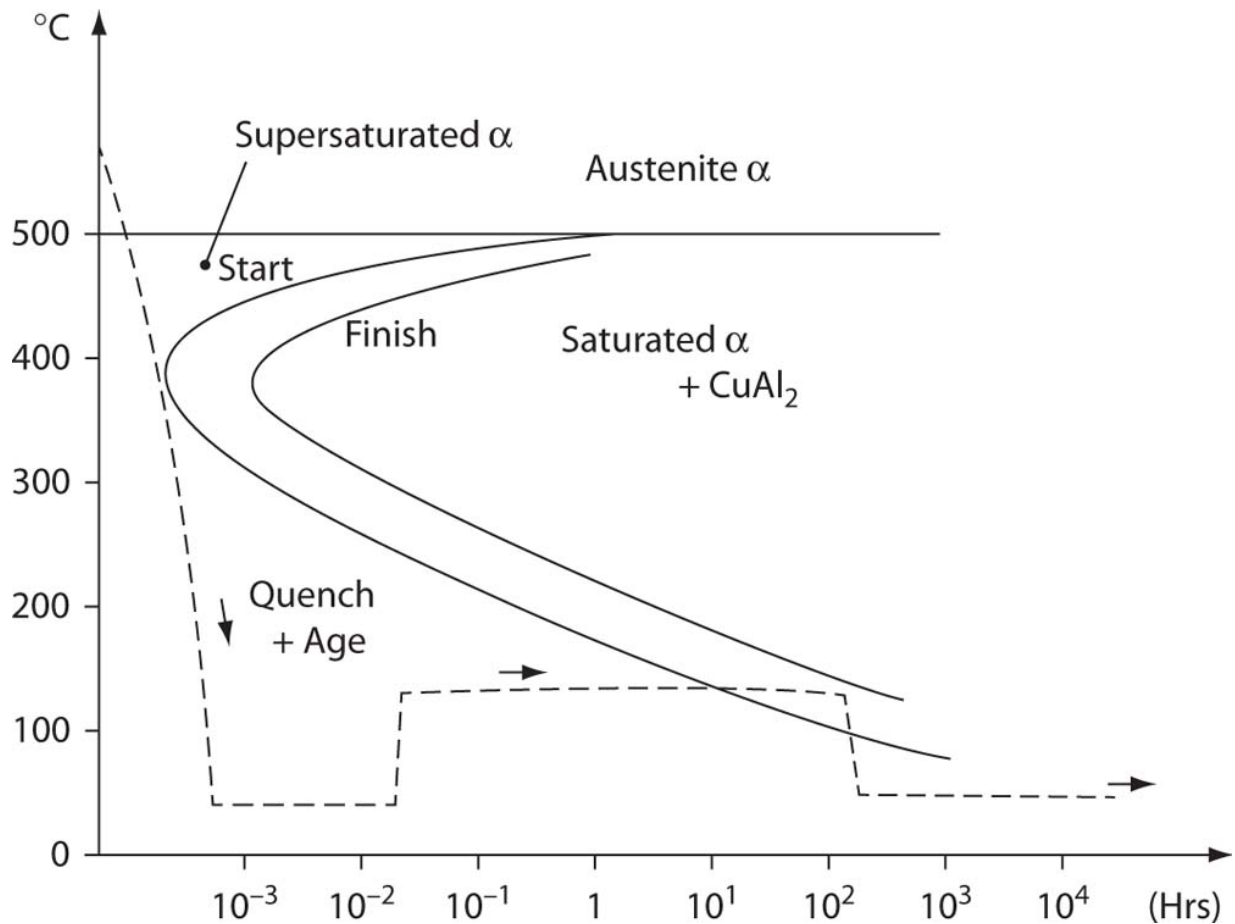


Figure 69 Cooling curve for Al-Cu alloy; 2000-series (typical).

Before examining the stages of hardening, it is appropriate to introduce the concept of the 'TTT dia-gram' (time-temperature-transformation), which also plays a pivotal part in the discussion on the hardenability of steels, albeit through an entirely different mechanism to that of precipitation hardening, as shown in [Figure 69](#). The TTT diagram enables the missing dimension of the equilibrium diagram, namely time, to be incorporated so that the stability of the phase(s) for a specific alloy can be plotted against temperature. Thus, a unique TTT curve is required for each particular composition, normally available from reference works as per the [Bibliography](#), and shown here for Al-4 per cent Cu. It enables the rate of cooling, and heating, to be established in order to determine how fast the quench need be to capture the high driving force of the severely super saturated solution, retaining the 4 per cent of Cu in

solid solution, where only 0.1 per cent Cu would obtain at equilibrium; later to precipitate as a fine dispersion of particles on ageing. Note that should the cooling rate be insufficient, the CuAl_2 phase would gather as an agglomeration of coarse particles forming along the Al grain boundaries, rather than the fine dispersion of CuAl_2 precipitates which later develop throughout the grain, as required for a fully realized precipitation hardening process.

It can be seen from the TTT diagram above, that the next stage is to heat-treat the alloy by holding at 150°C for the ageing process, for around 100 hours, in order to encounter the appropriate transformation range of the TTT diagram. Whilst this diagram shows an overview of the transformation, it does not provide the details of what is actually going on, which can only be discovered by the highly specialized work of the experimental scientist as this involves precipitates in the nanometre range, with particles from 1 to 10 nm ($1 \cdot 10^{-9}$ – $10 \cdot 10^{-9}\text{m}$) being critical sizes.

Returning to the hardening stages, the initial strength of $\sim 200 \text{ N/mm}^2$ suffers a small decline with the initial ageing, as this strength increase from the original annealed state is due to solid solution hardening, which would be lost entirely were it not for the intervention of the second stage in the process.

The second stage begins as a series of extremely small discs, around 10nm diameter and 1nm thick, emerge, 'precipitate', within the supersaturated phase, known as GP (Guinier Preston) zones. These GP zones, as shown below, are aligned, 'coherent', with the matrix, but their growth begins to exert a strain on the microstructure. This results in the initial rise in yield strength as dislocation movement through the matrix is made more difficult.

The third, final stage to peak hardness involves the growth of the GP zones to $\sim 100\text{nm}$ diameter and 10nm thickness, but still with the same coherency as before. This precipitate is known as ' θ ' phase, and leads to the greatest strength where the precipitates both directly obstruct the dislocation movement, and also generate a significant lattice strain due to their increased size, a further obstruction to dislocation flow. This all results in the 'bowing' of

dislocations around obstructing particles, where optimum particle spacing causes the dislocations to bow out like the bubble gum and wire mesh analogy, previously noted.

The fourth stage begins after the optimum formation of the θ'' particles, with subsequent growth to discs $\sim 1\mu\text{m}$ diameter and $1\mu\text{m}$ thickness, to become θ' phase. This is coarser than θ'' , and the previous optimum spacing of the particles increases to $>1\mu\text{m}$ thickness to become θ' phase. This over-ageing process would continue with particle coarsening and loss of strength, until eventually reaching the starting strength, albeit at a considerable time-scale of hundreds of hours.

Therefore, the TTT diagram enables prediction of the ageing process to reach optimum hardness, avoiding under- or over-ageing. This is an irreversible process, so that an over-aged item must be re-heat-treated, by returning to the original quenching from the α -phase at 550°C , and re-ageing.

A consequence of operating with this family of 2000-series alloys was encountered during the development of Concorde, as its normal operational temperature at supersonic speeds was around 150°C , and thus risked significant over-ageing. This resulted in the [introduction](#) of the RR 58 alloy, with a safe thermal stability at this service temperature. Note that the above is just an illustration of a common alloy, and heat treatment characteristics are specific to every alloy. Some alloys have been developed to age at room temperature for applications such as aircraft rivets, which are fitted immediately after holding at low temperature following the quenching, 'solution treatment' operation.

Work hardening

This is the final means for strengthening these alloys, and particularly important for the 1000-, 3000- and 5000-series aluminiums, as shown in [Table 35](#) below. These alloys are in the 'non heat treatable' class and this increase in strength is significant, compared to the annealed condition, in making them suitable for the

less critical structural applications. The mechanism of cold work is fundamentally the same as other alloy systems, with increasing dislocation density producing an effective barrier to flow and hence increased yield strength, although not to the levels achieved by the precipitation hardening (2000-, 6000- and 7000-series) alloys.

Table 35 Aluminium alloys – effect of work-hardening (0.2%PS N/mm²).

alloy	Fully annealed	Part-hard	Fully hard
1100	32	110	140
3005	64	130	190
5456	150	280	360

Methods of joining

The practical application in structures of the light alloys depends to a large extent on facility for joining. The 2000- and 7000-series aluminiums, whilst providing the highest strengths, are generally unweldable by the fusion process. This is evident in the manufacture of aircraft, which are still almost entirely joined by riveting, a process long-since abandoned in the manufacture of steel-based structures and vehicles. The 5000-series provides alloys capable of fusion welding without significant loss of strength. As described earlier, its strength is largely due to solid solution hardening. This means that fusion welding is suitable, using the TIG process with AC output to cope with the problematic, high melting point, oxide. It makes oxyacetylene welding a rather skilled process, necessitating a very aggressive flux to disperse the oxide, quite apart from the difficulties of control of the weld pool, and risk of general collapse. Electrodes, consumables, can be found in [Appendix II: Resources](#).

The 6000-series aluminiums are commonly joined by fusion welding, typically the 6061 alloy often used for cycle frames. However, in the as-welded state the HAZ properties are little better than the pre-aged condition, the material having suffered an accelerated over-ageing process, akin to that described previously. The only solution is to re-heat-treat the structure; this may be perfectly satisfactory, if expensive, provided that the effects of distortion are negated by restraining the item in a fixture. Whilst this is not necessarily a great problem for a cycle frame, where the joint is left as-welded, the potential benefits of this precipitation hardening alloy are not realized, and a 5000-series alloy should be considered.

Finally, the lower strength alloys such as 1000-series, are used to advantage where limited strength is a virtue, as in ring-pull drinks cans. Their great ductility permits deep drawing, and is also utilized in the compression joints formed by swaging a lap joint around the body to top.

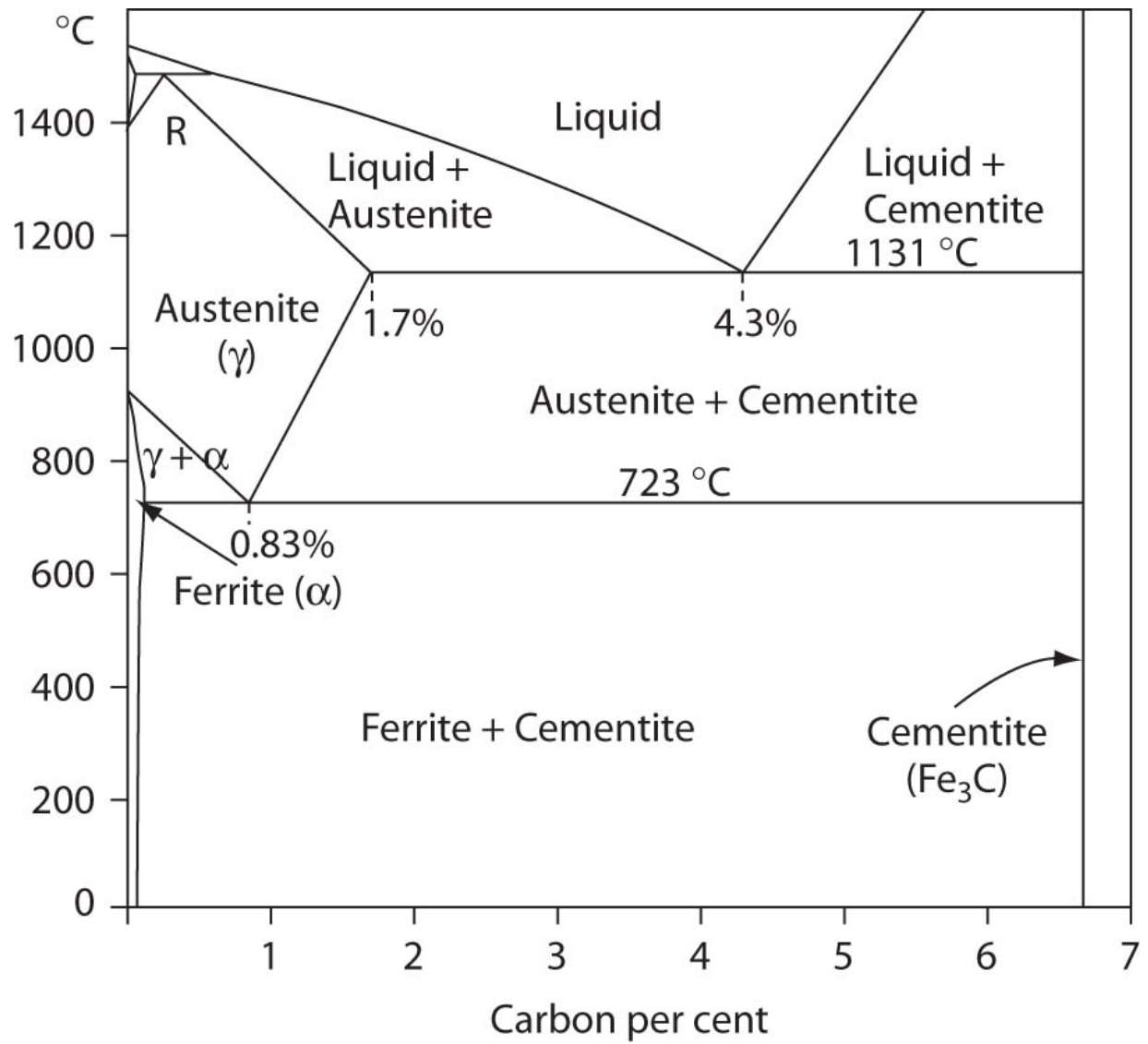


Figure 70 Fe-C equilibrium diagram – steel and cast iron (generally, steels <1.7%C, cast irons >1.7%C).

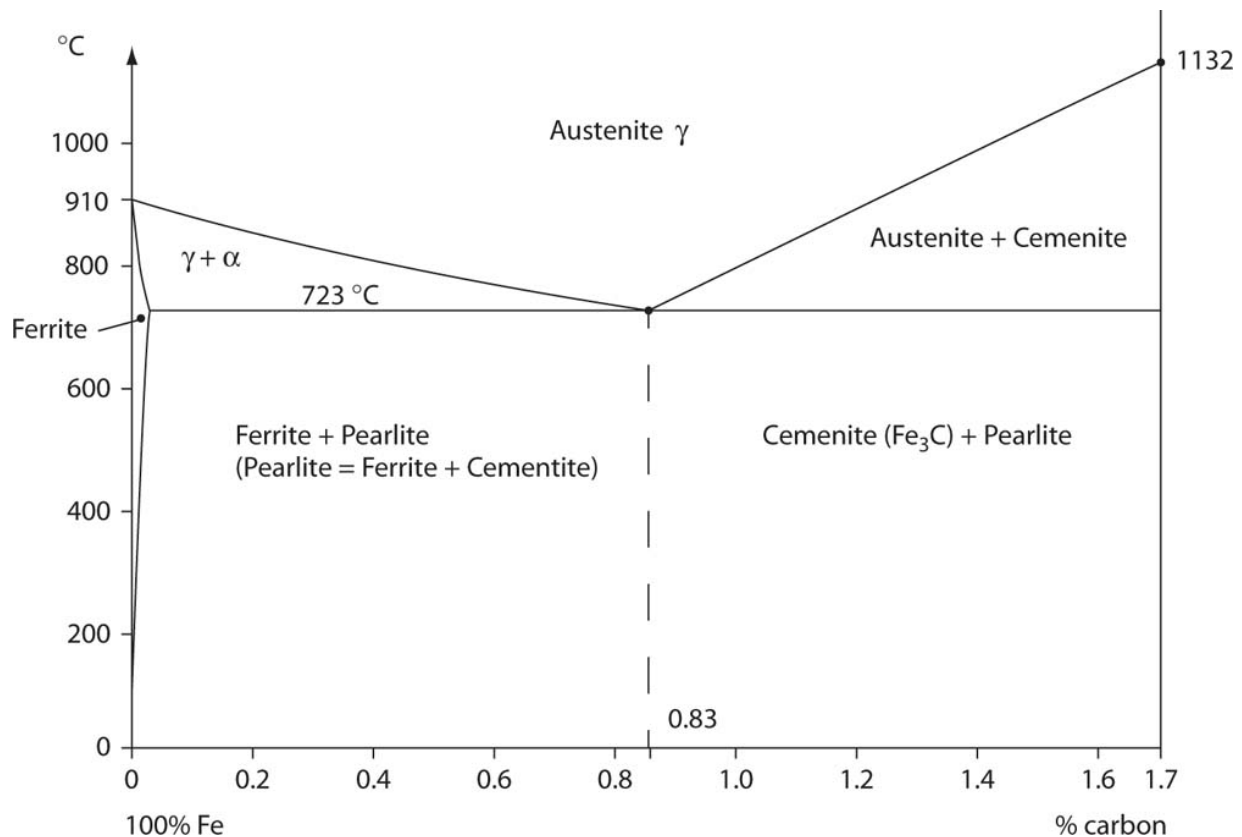


Figure 71 Fe–C equilibrium diagram – steel.

PLAIN CARBON STEELS

The next two sections re-visit the steels introduced in [Chapter 2](#). Plain C steels usually form a vital part in the background of the amateur worker, from basic model making to the most sophisticated experimental engineering but, despite a history of iron-working stretching back to antiquity, the iron carbon (FeC) and its derivatives should not be underestimated in their complexity at the fundamental level.

Taking the plain carbon steels, based on the equilibrium diagram approach, we are now in a position to understand something of the complexities that belie this familiar material, from its dead mild steel (<0.1 per cent C) to high C (1.2 per cent C) versions. [Figure 70](#) is an equilibrium diagram of the complete FeC system, illustrating the remarkable range, up to 7 per cent C, that encompasses the ‘steels’ (C <1.7 per cent) to the ‘cast irons’ (C >1.7 per cent). Although, for

the purposes of examining the steels, we confine ourselves to only the lower left hand corner, below C ~ 1.7 per cent and T $\sim 1000^{\circ}\text{C}$, for effects that take place at temperatures far below the liquidus of $\sim 1400^{\circ}\text{C}$. Note, however, that a discussion of fusion welding needs to recognize the solidification from temperatures well in excess of 1000°C . This is an area not normally the province of the worker in steel, un-involved in foundry work, as all of the following reactions and heat treatments extend no higher than the austenitic phase.

Whilst the high temperature equilibrium diagram illustrates a complex peritectic reaction at the liquidus, this need not concern us here as the material with which we work is the product of a well-developed manufacturing route discovered and developed at some considerable cost since the nineteenth century, by great metallurgical pioneers such as Bessemer, as described in the section on history in the [Introduction](#). We can safely assume that we are now working with steels of a metallurgically clean, homogenous structure, and of controlled compositions, from the start. This does not, however, mean that we cannot wreak havoc with the properties, such as excessive grain growth, or decarburization through prolonged overheating. Restoration of these properties is one of the features of this section.

Note that this is the case for steels from reputable sources dating from the late twentieth century, as the Markham Colliery illustration showed, for considerations of toughness in critical applications. And dramatic failures of gas pipelines and bridges, appearing on the Internet, indicate that the knowledge and experience of the modern steel maker may not have been applied as assiduously as one might expect in certain emerging industrial nations.

Before examining the detail of the steel section of the Fe–C equilibrium diagram ([Figure 71](#)), it is worth repeating the caution about the interpretation of equilibrium diagrams, particularly the fact that it is usual to examine the Fe–C diagram as it appears here, ignoring the fact that this is only a simplification of a practical alloy. As even plain C steels normally contain at least 0.3 per cent Mn, and other trace elements, clearly the simple Fe–C diagram should be treated with discretion, deferring to experience and established

values for temperatures of heat treatments. It will be noted that the boundaries of the austenitic phase rise steeply with carbon content, and there may appear to be a conflict between a temperature chosen from the diagram shown and that required by practice. The old adage, 'what works is right', should always prevail over a purely theoretical approach. The latter may be an over-simplification of complex behaviour, apparent as experience is gained.

The iron-carbon phases

From the Fe–C phase diagrams, we have the basic constituents (phases) of:

Ferrite (α – alpha phase)

This is almost pure iron, with only 0.006 per cent carbon at room temperature. 'Ferrite' is derived from the Latin 'ferrum', meaning iron. It is a soft and ductile phase, capable of accommodating substantial cold work, and strongly magnetic at room temperature. A bcc structure, of atomic spacing $a = 2.861 \text{ nm}$, is shown in [Figure 72](#).



Photo 115 Butane gas heating torch, lower heat output but can be more manageable than oxy-acetylene for soldering.



Photo 116 Oxy-acetylene torch with 'pepperpot' (multi jet) nozzle for more general heating than with a welding jet.



Photo 117 Oxy-acetylene torch held ready for hot working of bar material.



Photo 118 Application of multiple heat sources for working larger material.



Photo 119 Direct heating by gas torch on firebricks.



Photo 120 Material at lower bound for hot forming (previously overheated).



Photo 121 Material at higher bound for hot working, can lead to excessive grain growth if overdone.



Photo 122 Gross overheating of material, with rapid surface emissions and metallurgical damage.

Austenite (γ – gamma phase)

A solid solution of carbon in iron, which can accommodate up to 1.7 per cent carbon at $\sim 1150^{\circ}\text{C}$. Named after its discoverer Sir William Austen, it is also soft and ductile, but rather less so than ferrite. In contrast to ferrite it has an fcc structure with $a = 3.548\text{ nm}$ and, as we see later, can accommodate a carbon atom, being substantially smaller, on the b axis between the outer array of its atoms. This is a crucial feature of all the conventional hardening facilities that may be bestowed upon the steels – a simple truth that again belies a complexity of micro-mechanisms en-route to a strong and tough end result. Austenite is basically non-magnetic, existing above the ' A_1 ', (723°C) line on the phase diagram.

Cementite (Fe_3C – 'iron carbide')

An 'intermetallic' compound of iron with 6.7 per cent C, of great hardness and low ductility. Its name derives from the 'cementation' process, the laborious method applied to re-introducing only sufficient carbon back into the bars of ferrite employed as wrought iron, often called 'steel' prior to the mid-nineteenth century; after this wrought iron had been produced by removal of the excess carbon in cast iron! Magnetic at room temperature, it changes to a para-magnetic state above 210°C .

Pearlite

Pearlite (and Fe_3C) is sometimes erroneously referred to as a phase, which it is not, rather a combination of the two separate components of ferrite and cementite, forming a lamellar structure in the manner of a plywood with sheets of grossly different strength. Its name is derived from its attractive 'mother-of-pearl' appearance, readily observable in the optical microscope, normally from magnifications of 100 times upwards. Containing ~ 88 per cent ferrite and 12 per cent cementite, it is produced when austenite is cooled

at a slow rate, below the A_{R1} line of 723°C , as per the equilibrium diagram. It produces a hardness, as measured with a Rockwell hardness tester, of ~ 20 to $30 H_{RC}$.

Ledburite

This is the other constituent of the FeC diagram, a mixture of austenite and cementite, but as this forms above the 1150°C line it does not normally form part of our concern with the heat treatment of steels in the solid state.

Features of the FeC phase diagram

Certain characteristics of steels and their heat treatment, relating to particular aspects of the FeC diagram, have a habit of recurring. These include important lines such as the A_1 at $\sim 723^{\circ}\text{C}$, with the additional subscript of A_{R1} indicating cooling, as there is often a small difference from the equilibrium value of the diagram. The A_3 and A_{RM} lines are important as the boundaries between austenite, and ferrite plus cementite, before and after the 'eutectoid' point at 0.8 per cent carbon, respectively. Illustrated below are the microstructures observed in the optical microscope. Note that the actual appearance is a function of, among other things, the etchant – Nital (5 per cent nitric acid in water) being commonly used – and magnification. The relative proportions of the microstructural constituents, coupled with a slow cooling regime, can provide the experienced metallurgist with a fairly accurate guide to the chemical composition, and thus a good idea about suitability for intended service, and compliance with specification. It is only necessary here to have a general picture of the sorts of structures expected from given compositions. Considerable experience is required for interpretation of microstructures, without the back-up of chemical composition and mechanical testing, for which the services of the commercial test house or academic institution will probably be required.

HEAT TREATMENT OF STEELS

Steel heat treatments may be categorized as follows:

- ◆ Annealing
- ◆ Normalizing
- ◆ Hardening: conventional
- ◆ Tempering
- ◆ Hardening: special processes
- ◆ Alloy steels and transformation diagrams
- ◆ Case hardening

Annealing

There are a number of special types of annealing although they can sensibly be prioritized according to application:

- ◆ Hardness reduction
- ◆ Machinability improvement
- ◆ Internal (residual) stress relief
- ◆ Precursor to cold work
- ◆ Microstructural improvement (re-crystallization)

All the above utilize the same principle of steady heating then holding at the designated temperature, as a rule of thumb for one hour per inch (25mm) of thickness, with a minimum of 30 mins. This is followed by as slow a cooling rate as possible, normally simply in the furnace with power off from the holding temperature. For a small muffle furnace the natural cooling rate may be sufficient (although my furnace's cooling rate is an order of magnitude too fast and residual power is required) for the desired rate of 30–50°C per hour, at least down to 500°C after which natural furnace cooling is perfectly satisfactory, with air cooling from 300°C. Alloy steels need very careful control of cooling rates due to their transformation characteristics, of which more later.

Table 36 Carbon steels – hardness and fullannealing temperatures.

Carbon (%)	Full-annealing temp. (°C)	Hardness (HV)
0.18–0.22	860–900	115–155
0.23–0.29	850–890	135–185
0.30–0.39	840–880	145–210
0.40–0.54	820–870	155–220
0.55–0.79	800–840	165–235
0.80–1.0	780–830	175–240

Table 36 indicates approximate annealing temperatures as a function of carbon content, with a guide to the hardness as measured with the Vickers hardness tester. This shows the expected range of hardness simply due to the carbon effect on the lattice strain, producing an almost linear relationship. These holding temperatures and cooling rates apply to the most rigorous condition, providing the lowest hardness, known as a ‘full annealing’ treatment.

Ideally, this treatment should be undertaken in a controlled-atmosphere furnace in order to avoid any surface oxidation, but this is often not possible so either an allowance for cleaning off the surface scaling, or protecting with a suitable compound such as a chalk wash, is required in a full annealing treatment.

A similar process will normally provide a machinability improvement, although for mild steel a normalizing treatment may be preferred for a fine, tear-free cut in the lathe. Small variations of the holding time and cooling rate, best optimized by experiment, can significantly affect these properties.

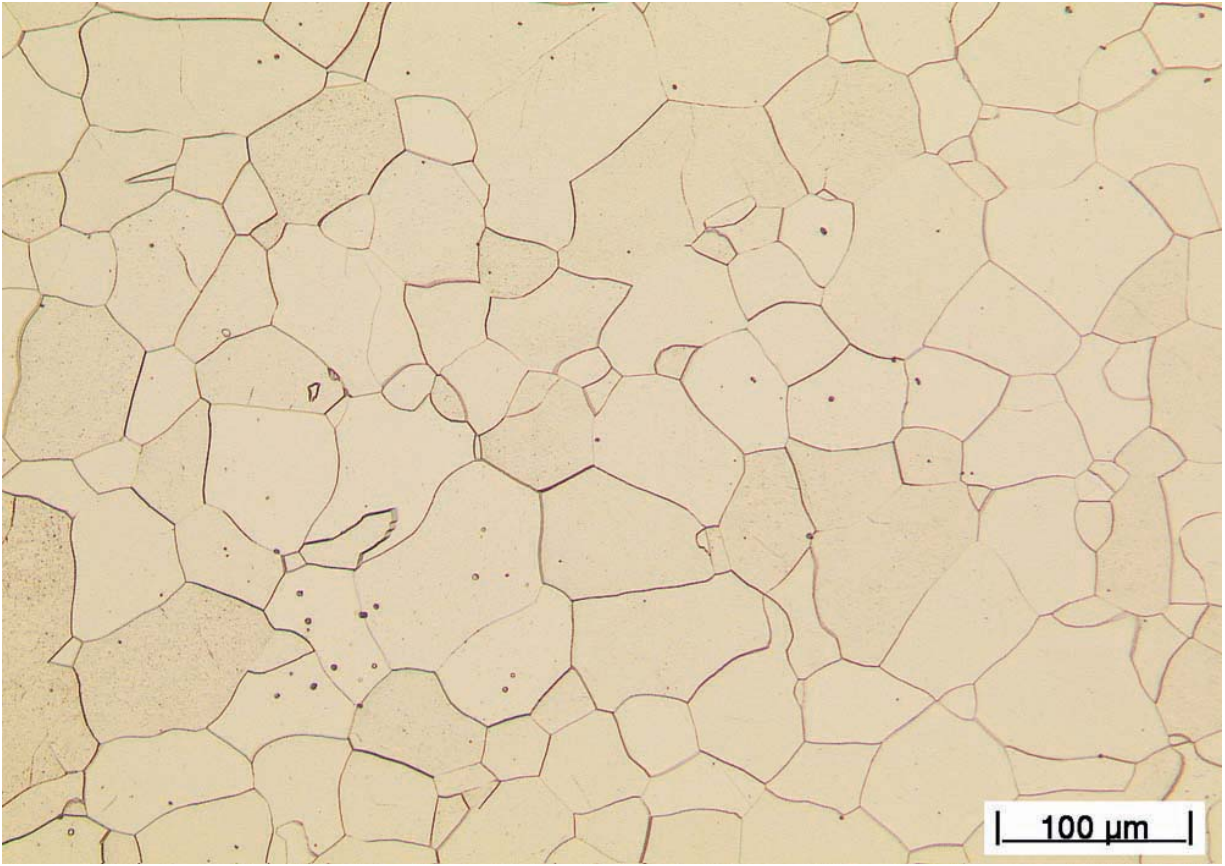


Photo 123 Steel microstructure × 200; fully ferritic, low carbon.

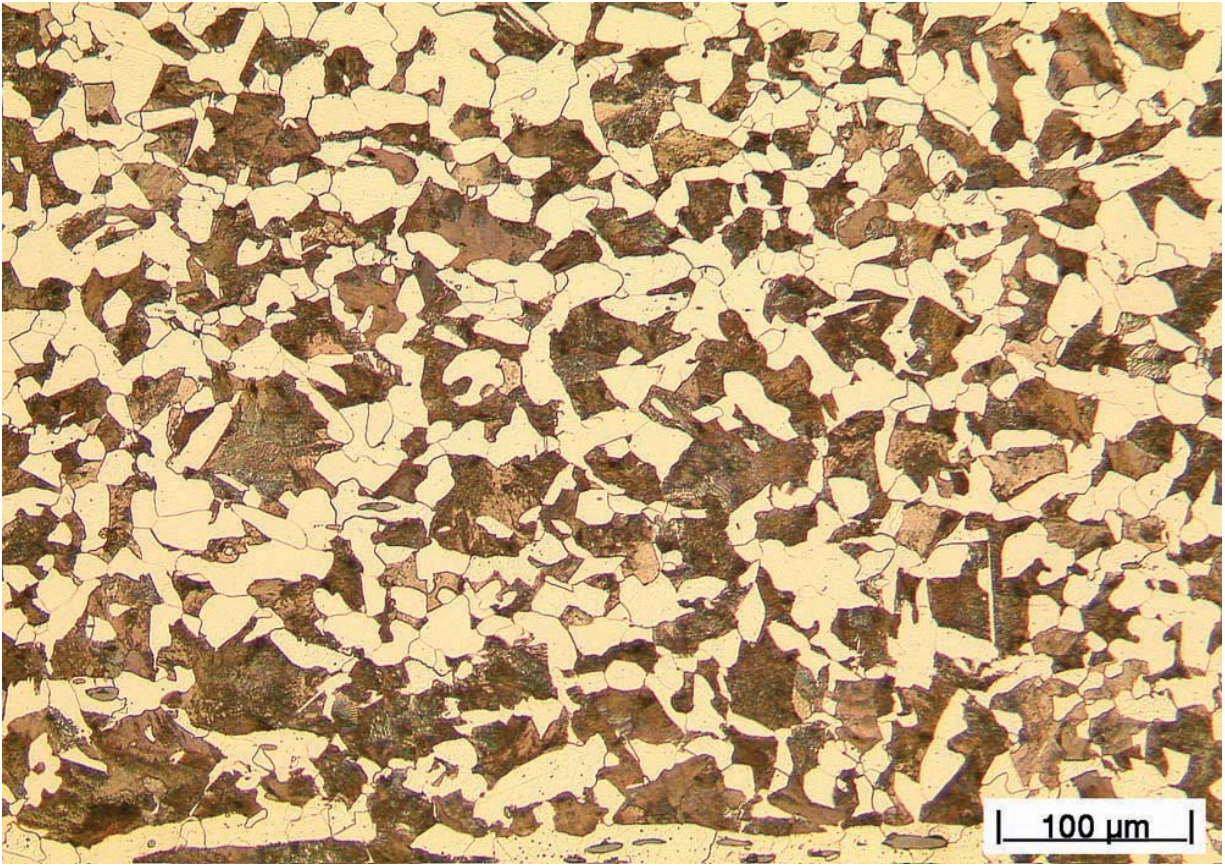


Photo 124 Steel microstructure $\times 200$; ferrite and pearlite of 0.4% carbon steel.

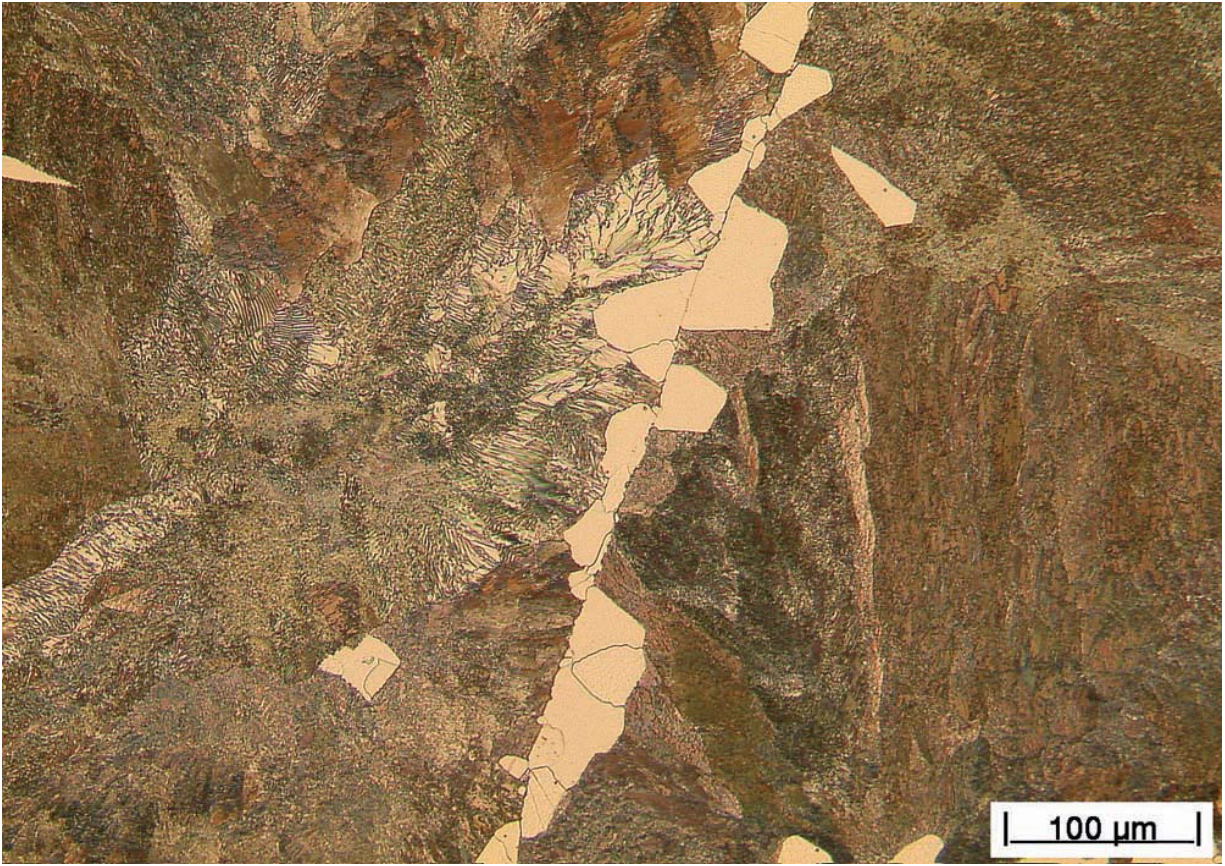


Photo 125 Steel microstructure $\times 200$; ferrite and pearlite of 0.7% (hypo-eutectoid) carbon steel.

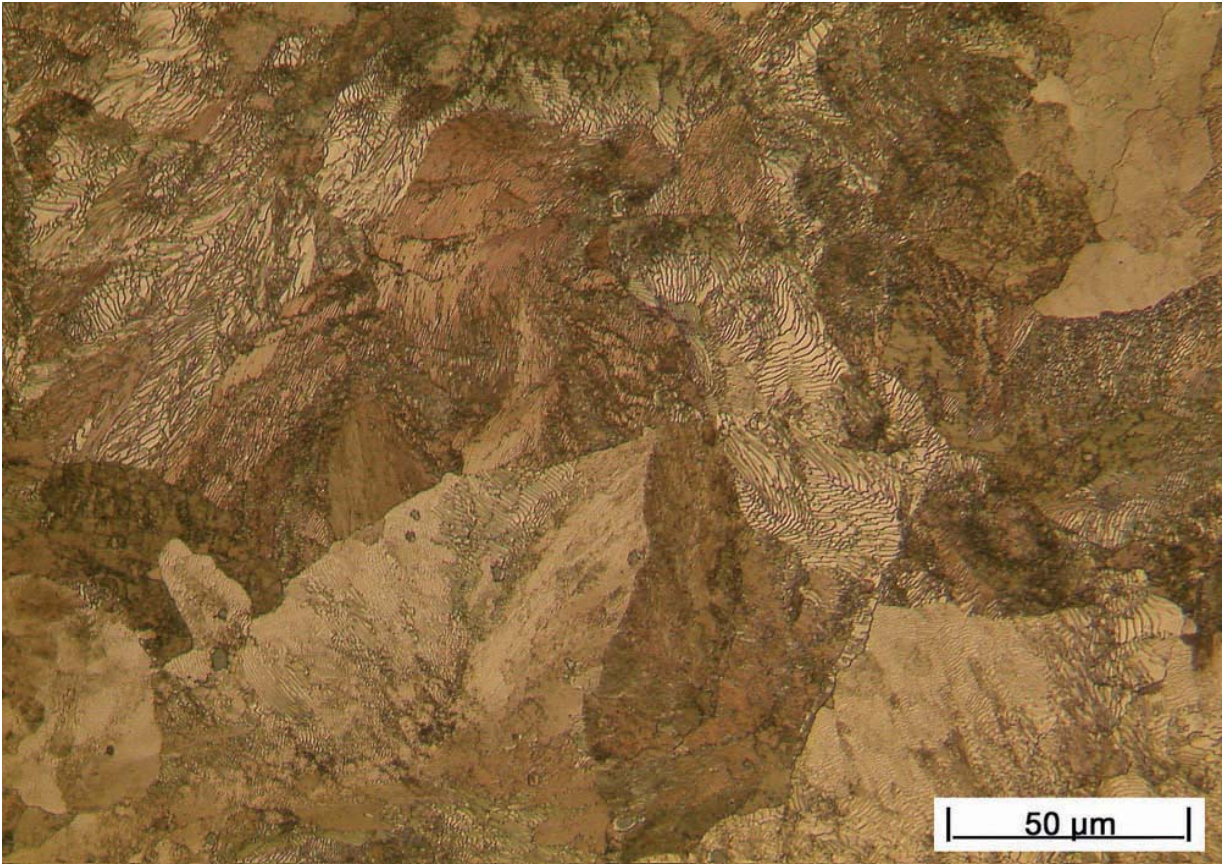


Photo 126 Steel microstructure $\times 500$; fully pearlitic 0.8% (eutectoid) carbon steel.

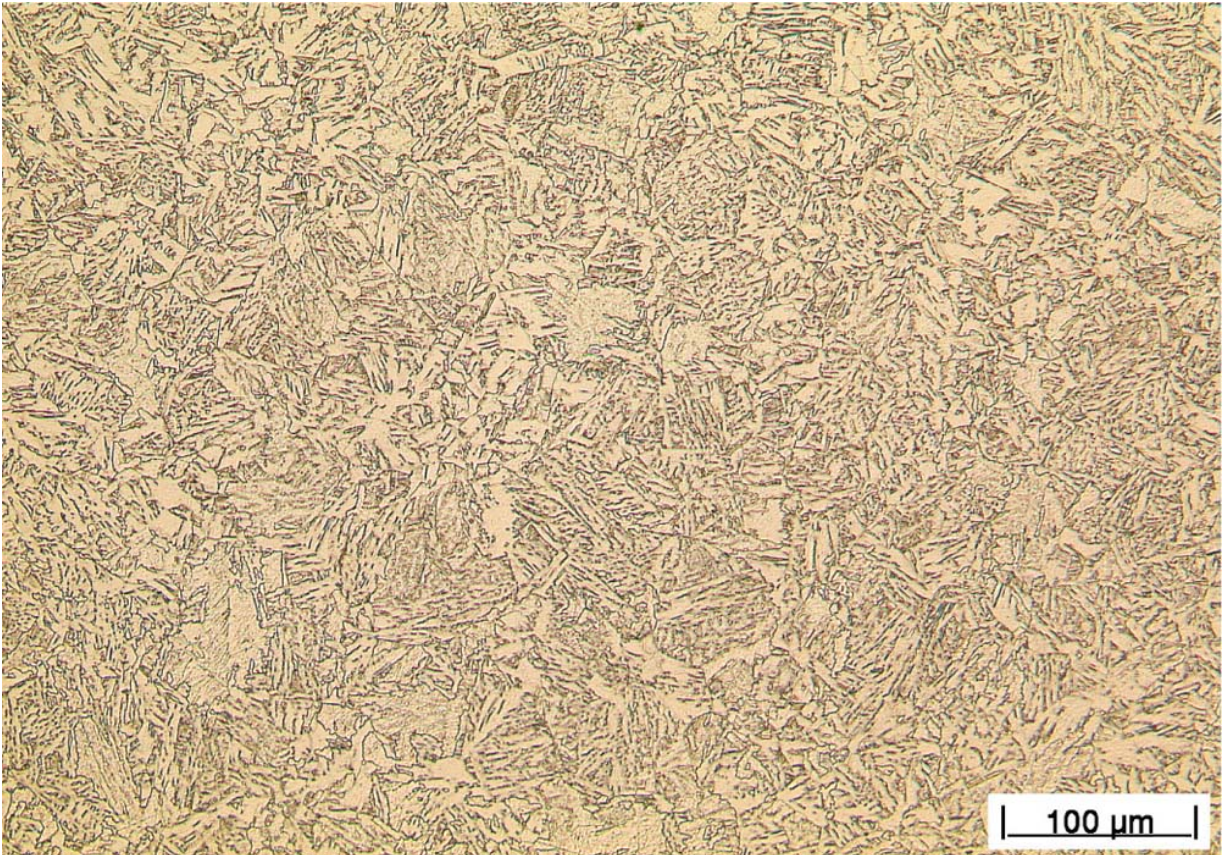


Photo 127 Steel microstructure $\times 200$; ~ fully transformed bainitic phase of a high carbon steel.

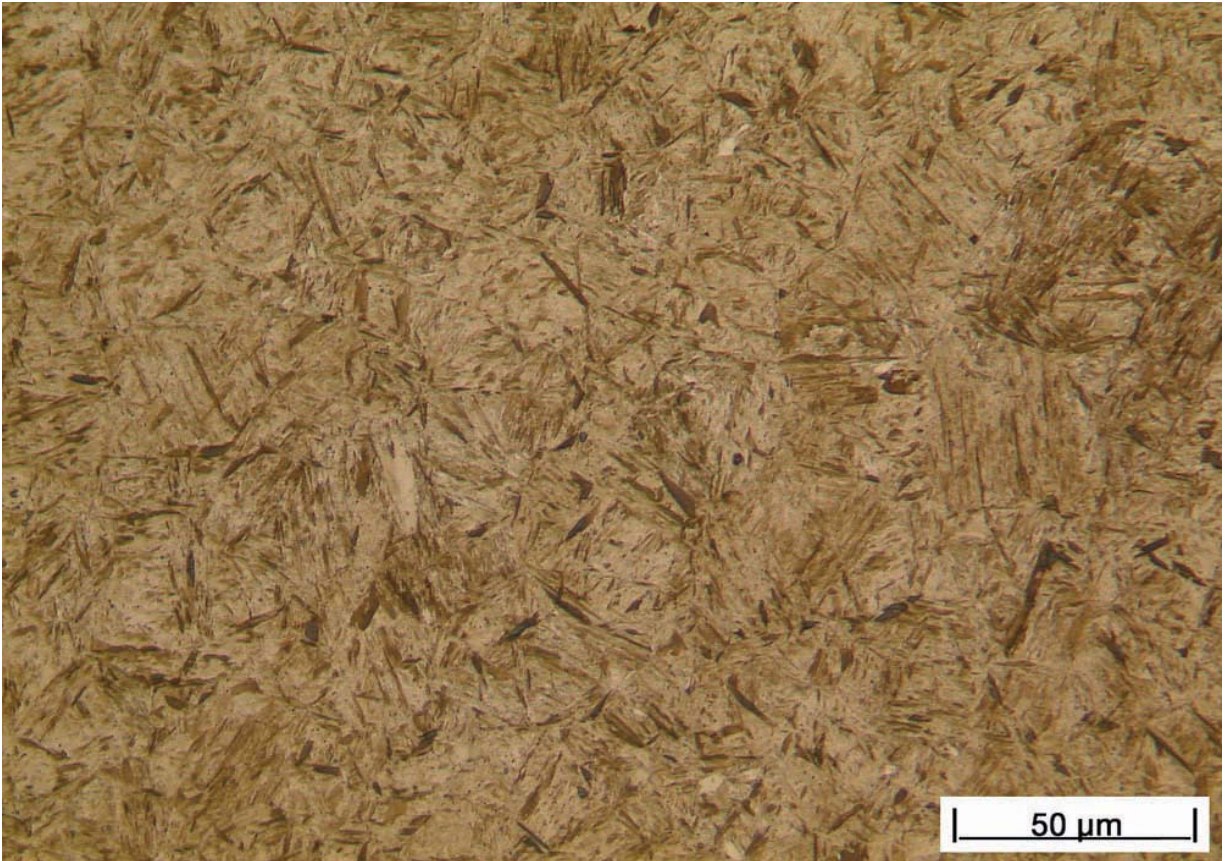


Photo 128 Steel microstructure $\times 500$; ~ fully martensitic phase, high carbon steel. (Photo: Tata Steel)

If stress relief and removal of residual stress is the principal requirement, then the holding temperature can often be reduced to 600°C . This is the normal temperature for post-weld heat treatment (PWHT), and satisfactory results may even be produced at 500°C , depending on circumstances.

An annealing process known as 'spheroidizing' may be applied to the higher-carbon steels, tool steels, and some alloy steels, by heating to $\sim 700^{\circ}\text{C}$, or just above the A_{C1} , lower critical line. This promotes the spheroidization of the pro-eutectoid cementite from the pearlite, followed by holding below the A_1 . This can also be used to allow further cold work on certain drawn steel tubes, such as used for cycle frames.

Intermediate annealing is often used at intervals during an extensive cold working process on lower carbon steels, and can

otherwise be known as re-crystallization annealing, conducted in a similar manner to the above but with a holding temperature below A_{C1} .

Re-crystallization annealing, as explored earlier in this chapter, follows a similar pattern. This requires a holding temperature below A_{C1} , in the range of 500–700°C, depending on the re-crystallization temperature of the particular steel. This illustrates why hot working at red-heat, well above the re-crystallization temperature, permits continuing deformation without strain hardening. Metals with a re-crystallization temperature below room temperature, as previously noted for lead, allow extensive compressive deformation at room temperature without work hardening and the need for heat treatment.

A very useful further application of annealing is for castings, permitting relaxation of the process stresses, and an improvement of microstructure that may render large castings suitable for service in competition with wrought or fabricated product. Although likely to be of less direct importance to the home worker, this explains how castings can safely be processed for critical applications, such as vehicle wheel hubs.

Normalizing

Whilst there appear to be many similarities between annealing and normalizing, in the latter the holding temperature for full normalizing can be some 20–50°C higher and, critically, the cooling rate is faster. This derives from air cooling, after the same soaking period, by removal from the furnace. It finds common application in industry, and material is often specified as being normalized as-supplied. It is not only a more economical process with greatly reduced cooling times, the resulting microstructure provides a generally finer structure of pearlite, with a slightly higher strength than that produced by annealing, and by avoiding excessive ductility it helps to optimize the machinability of mild steels.

Table 37 provides a guide to normalizing treatments and properties, in comparison with that for annealing. Similarly for castings, normalizing may be used in lieu of the inoculation control described earlier in this chapter, to transform the typical as-cast structure to the more uniformly equiaxed condition.

Note that whilst air cooling is fine for the lower carbon plain steels, this may not be the case for higher carbon and low alloy steels of smaller sections. For these, attention must be paid to the effects of cooling rate, as explored in more detail with the TTT or CCT diagrams, often requiring slow furnace cooling.

Table 37 Carbon steels – hardness and normalizing temperatures.

Carbon (%)	Full-annealing temp. (°C)	Hardness (HV)
0.18–0.22	900–925	125–165
0.23–0.29	890–910	145–195
0.30–0.39	880–900	155–225
0.40–0.54	840–890	185–235
0.55–0.79	810–840	215–285
0.80–1.0	800–840	270–320

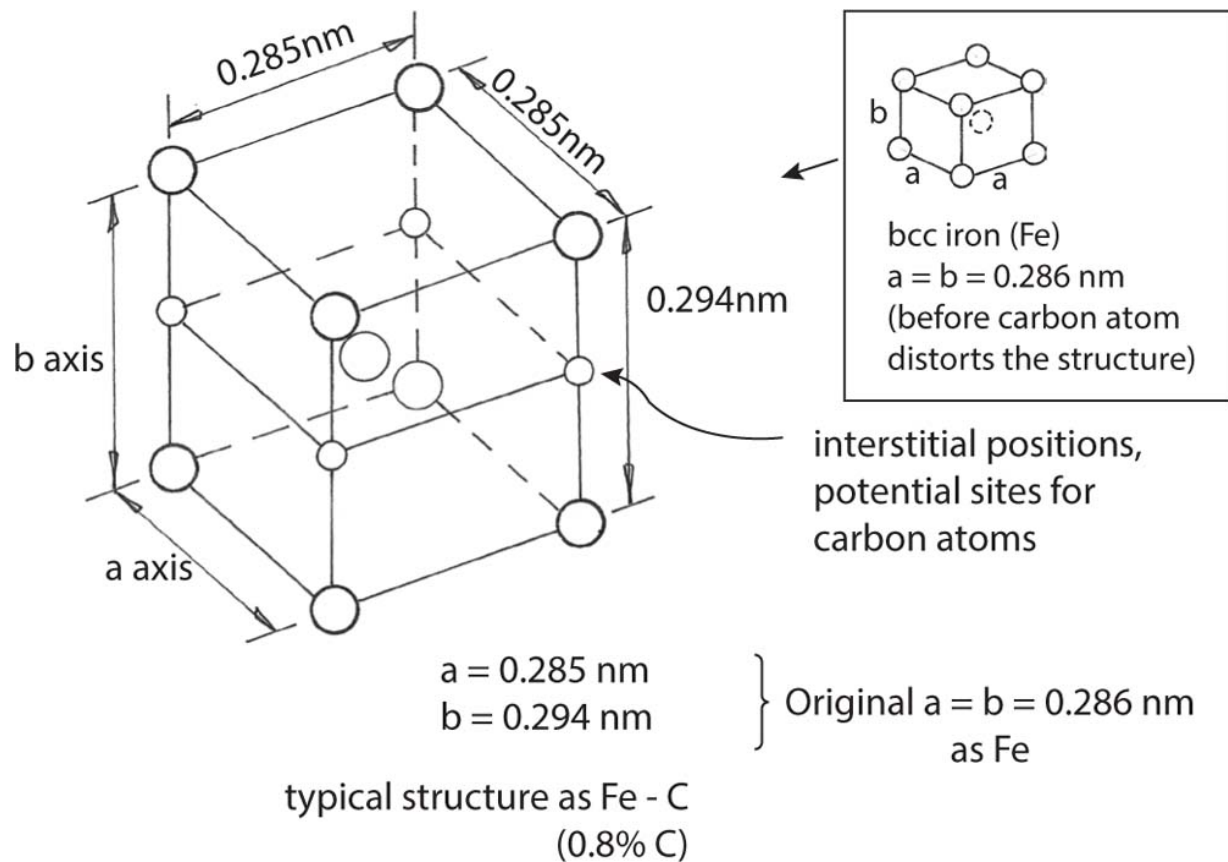
Hardening (conventional)

On the one hand, hardening of steels couldn't be simpler: just rapid cool from red-heat and we have produced a very hard steel. On the other hand, however, this apparently simple process once again belies a number of complex and fascinating forms of behaviour that have only been comprehensively unravelled in the latter half of the twentieth century.

As noted in the [introduction](#) to this book, the history of strengthening of steels goes back to antiquity but the physics behind

it was beyond understanding until the relatively recent development of the scientific method, and finally the means to examine transformation products down to the nanometre scale with electron microscopy. In place of scientific theory, earlier ages relied on supposition and myth, such as the often grizzly notions of the Greeks and Romans who seemed to have a penchant for quenching their steels into the blood of animals or hapless slaves. Less appalling are the amusing stories of the nineteenth-century Yorkshireman (a tale promulgated by Lancastrians, some might suspect...) who made a highly profitable business exporting to Japan barrels of Sheffield water reckoned to possess 'magical' properties for quenching steel, the success of which was simply due to the skill and experience of the indigenous steelmakers.

The basic principle of hardening of steel centres around the somewhat remarkable effect of the carbon atom which can be accommodated interstitially on the 'b' axis of the austenitic fcc crystal, as shown in [Figure 72](#). This carbon atom, bearing in mind its scarcity in the matrix, there being only around one carbon atom for every ten in the austenite of a eutectoid (0.8 per cent C) steel, possesses great mobility at the cooling rates associated with annealing and normalizing. This enables the room temperature constituents of a moderately cooled steel to achieve the pearlitic structures indicated in the Fe–C equilibrium diagram, [Figure 71](#).



Change from Fe structure to Fe - C crystal as martensite, from bcc (body centred cubic) to fct (face centred tetragonal)

Figure 72 Crystal structure – Eutectoid ($\sim 0.89^\circ$) steel Martensite.

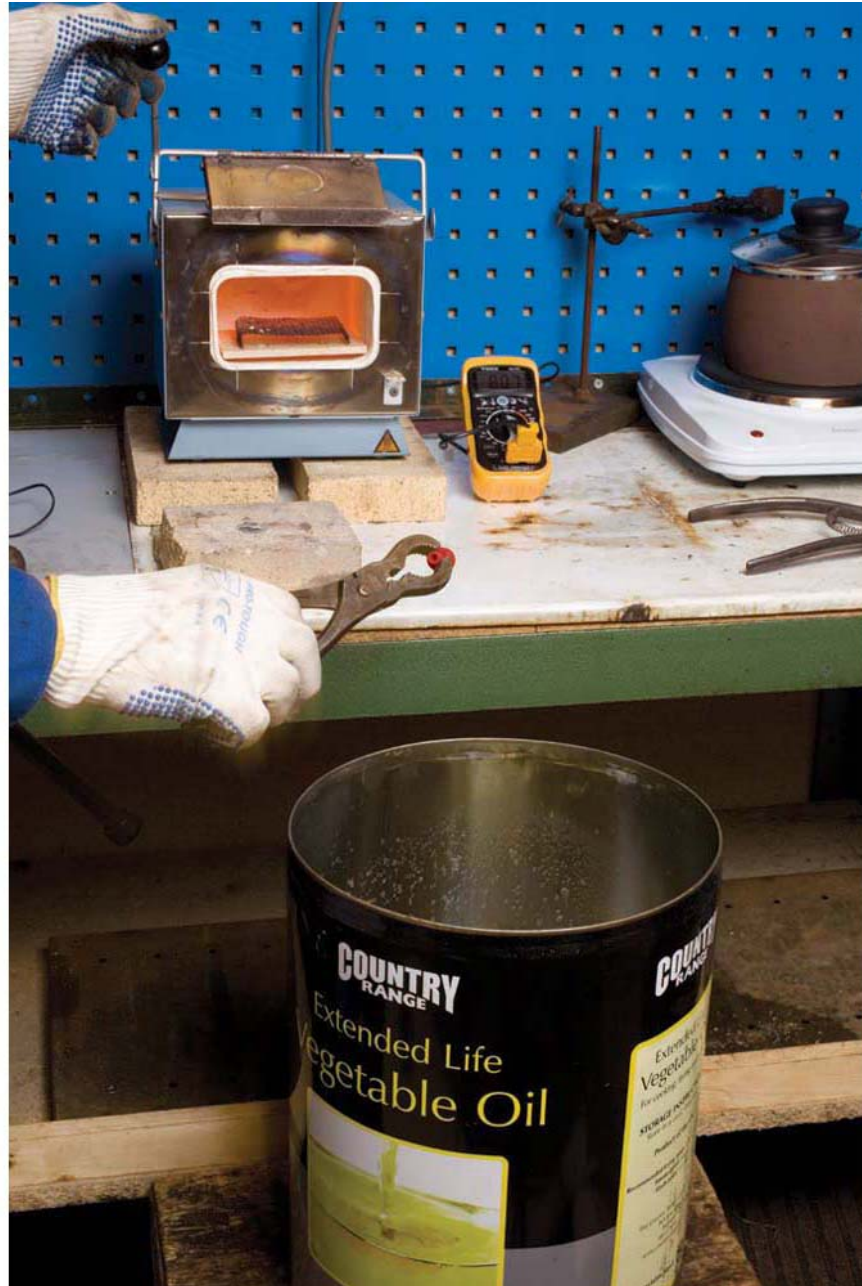


Photo 129 Removal of steel testpiece for quenching into water.

However, the heat treatments met so far are used to produce an improvement of ductility and toughness, but with a reduction in hardness and strength. In order to produce the strong steels demanded for the applications discussed throughout the book, we must turn to hardening processes. Conventionally, these rely on heating the material to some 20–50°C above the upper critical point,

to obtain a fully austenitic structure, holding to stabilize, then rapid quenching. The quench rate is determined from the TTT or CCT diagram and obtained by choice of the quench medium and method, as described later.

Whilst apparently simple, this treatment is by no means foolproof, and a reasonable understanding of the basic physics behind the transformations that provide this spectacular result is useful, to cope with tests that fail to achieve the desired result, as well as establishing methods to circumvent problems such as quench-cracking, or early service failures.

Instead of the slow cooling and transformation by the normal *diffusion* process that the mobility of the carbon atom so readily accommodates, if the cooling rate is increased by quenching into water or oil then the fcc austenitic phase effectively traps the carbon atom held interstitially on the b-axis, without time to diffuse out to the equilibrium product, pearlite. The effect in steels is for this stressed crystal structure to be forced toward the bcc equilibrium structure, which it aims for at room temperature. This driving force results in the transformation into a body-centred tetragonal (bct) structure, the essence of the hardening process in steels. The phase produced is a 'martensitic' one, a generalized term for a transformation product achieved by a *non-diffusion* or *displacive* reaction. It produces some very special characteristics and, despite its somewhat complex nature, is worthy of some further exploration.



Photo 130 Immersion of testpiece for rapid quench into water.

Martensitic reactions

The essence of the martensitic reaction is that, in the case of the FeC diagram, the austenitic, fcc crystal structure is arrested as a supersaturated solution in this high temperature phase. Then, on rapid cooling, the lattice strains associated with the unrequited desire to transform into the bcc structure, provide the very large

driving force to trigger off the displacive, diffusionless, ejection of platelets of martensite from the austenitic grain boundaries. This is a fundamentally different process compared to that previously met in the *diffusion* processes that occur at slower than the critical cooling rate for martensite. Martensite growth proceeds at an extremely high velocity, around the speed of sound in iron, some 5000 m/s – almost instantaneously. However, it only proceeds until the strain energy, available at the precise temperature, is expended. Initiation of this transformation is designated ' M_s ' (martensite start), and is completed at ' M_f ' (martensite finish), so that interruption at a temperature between M_s and M_f on the critical cooling rate graph causes a halt in the growth of martensite. The range of M_s to M_f is often as great as 200°C, and can begin and end almost anywhere from 400°C to deeply sub-zero temperatures, depending on the particular composition.

Another feature of martensite in the M_s and M_f values, which are entirely fixed by the chemical composition, is that it will occur providing cooling proceeds at any rate equal to, or faster than, the critical rate, as shown on the TTT, or more accurately the CCT diagrams, as shown below. The speed of the martensite plate growth is similar to that for dislocation movement, as it progresses by the extremely rapid making-and-breaking of atomic bonds. This occurs as a precise, 'military', atomic movement is transferred along the growing martensite disc. In fact, these grow as 'discs' or 'plates', but observed in two dimensions in the optical microscope appear as characteristic 'needles', spreading right across the austenite grains as the M_f is approached. In practice, full transformation is often not achieved, as seen in [Figure 73](#), where M_s is well above room temperature and M_f is not attained until sub-zero temperatures are reached. The volume transformed is therefore a simple function of progress from M_s to M_f , as also shown below.

As there is no diffusion of atoms in this displacive process, unlike for the former, there can be no *compositional* change. However, a

noticeable *volume* change, in the order of a 5 per cent increase, is typical for martensite in steel.

Finally, there is a strict coherence between the new martensitic phase and the original, parent, lattice. This interface is known as a 'habit' plane, where the distortion between the parent and new phase is accommodated by a kink, and strain, as shown in the martensite 'lens', shown in [Figure 74](#).

Returning to the 0.8 per cent C, eutectoid steel, the martensite has produced the body-centred tetragonal structure shown in [Figure 72](#), the consequence of not quite achieving the final change from fcc to the equilibrium bcc structure. This lengthening of b axis, and contraction of a axis, provides a strain of great importance to each affected grain, effectively blocking dislocation movement so effectively that the hardness of a fully transformed martensitic steel structure can achieve a five-fold increase, compared to pearlite. This increase can be so great as to produce a microstructure of inadequate toughness; hence the need to undertake a 'tempering' process in order to improve toughness at the cost of a limited strength reduction.

Non-steel martensites

Before leaving the subject, it is worth reiterating that 'martensite' is only a generic term for the diffusionless phase change, characterized by rapidity of growth, coherence with the existing structure, and zero compositional change – all of which sharply contrast with the other, diffusion-led, phase changes.

Whilst the FeC martensites are by far the most important in engineering, many other systems, even non-metals, exhibit martensitic reactions. Although these are of only peripheral interest, they do sometimes display some rather unexpected properties and have led to the discovery of the curious 'memory alloys'. These, unlike steel martensite, utilize the perfect *reversibility* of their martensite to 'recover' an original shape, prior to deformation in its martensitic phase, below M_f . The original shape is recovered simply

by increasing temperature to above M_s until it has been transformed back to its original high temperature phase. This was first discovered in the 1930s, and in the USA in the 1960s it was accidentally found during a laboratory management meeting(!) in a 50/50 alloy of nickel and titanium, to become known as 'Nitinol', (from Nickel Titanium Naval Ordnance Laboratories).

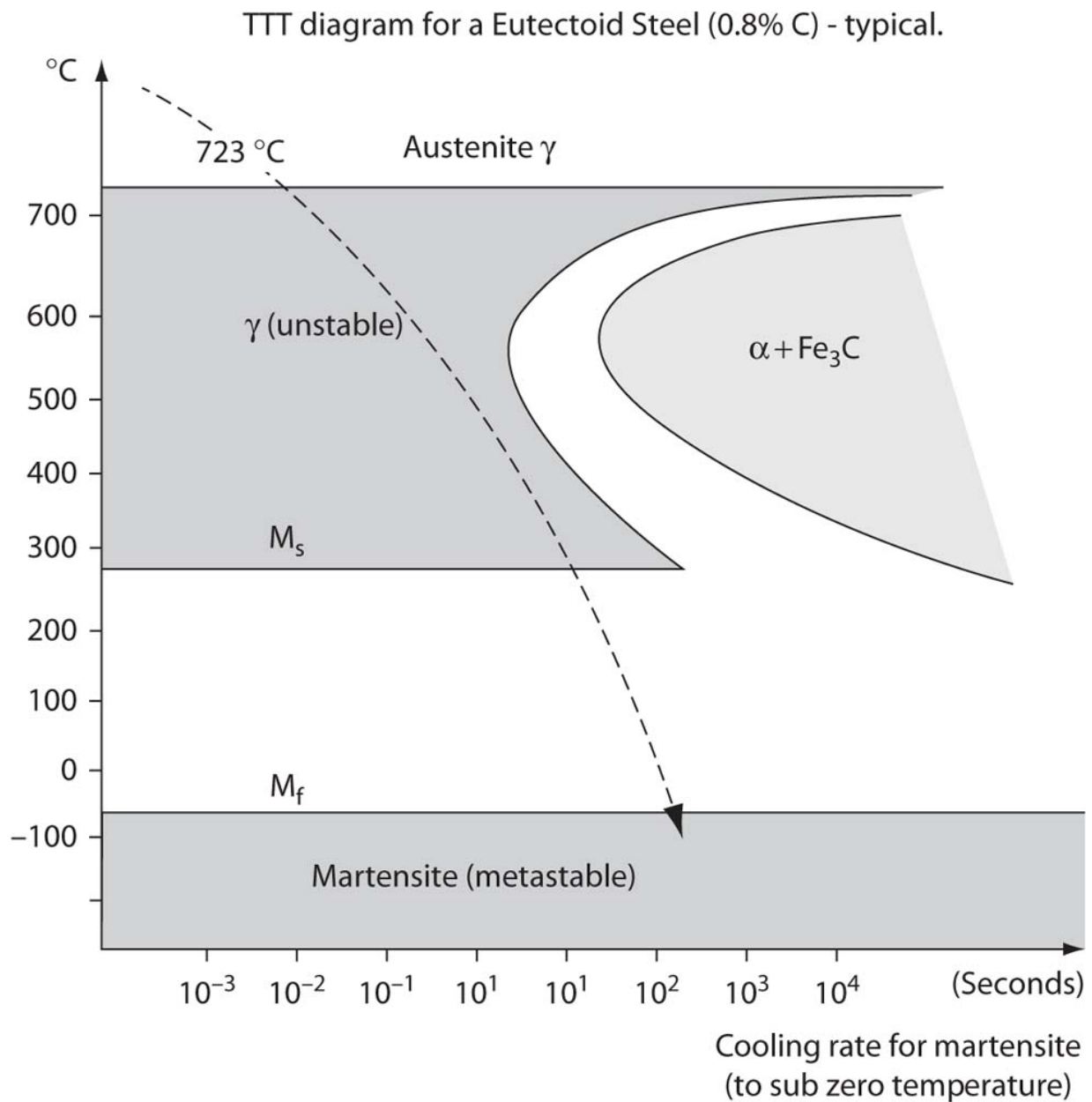


Figure 73 TTT diagram for a Eutectoid steel (0.8%C) – typical.

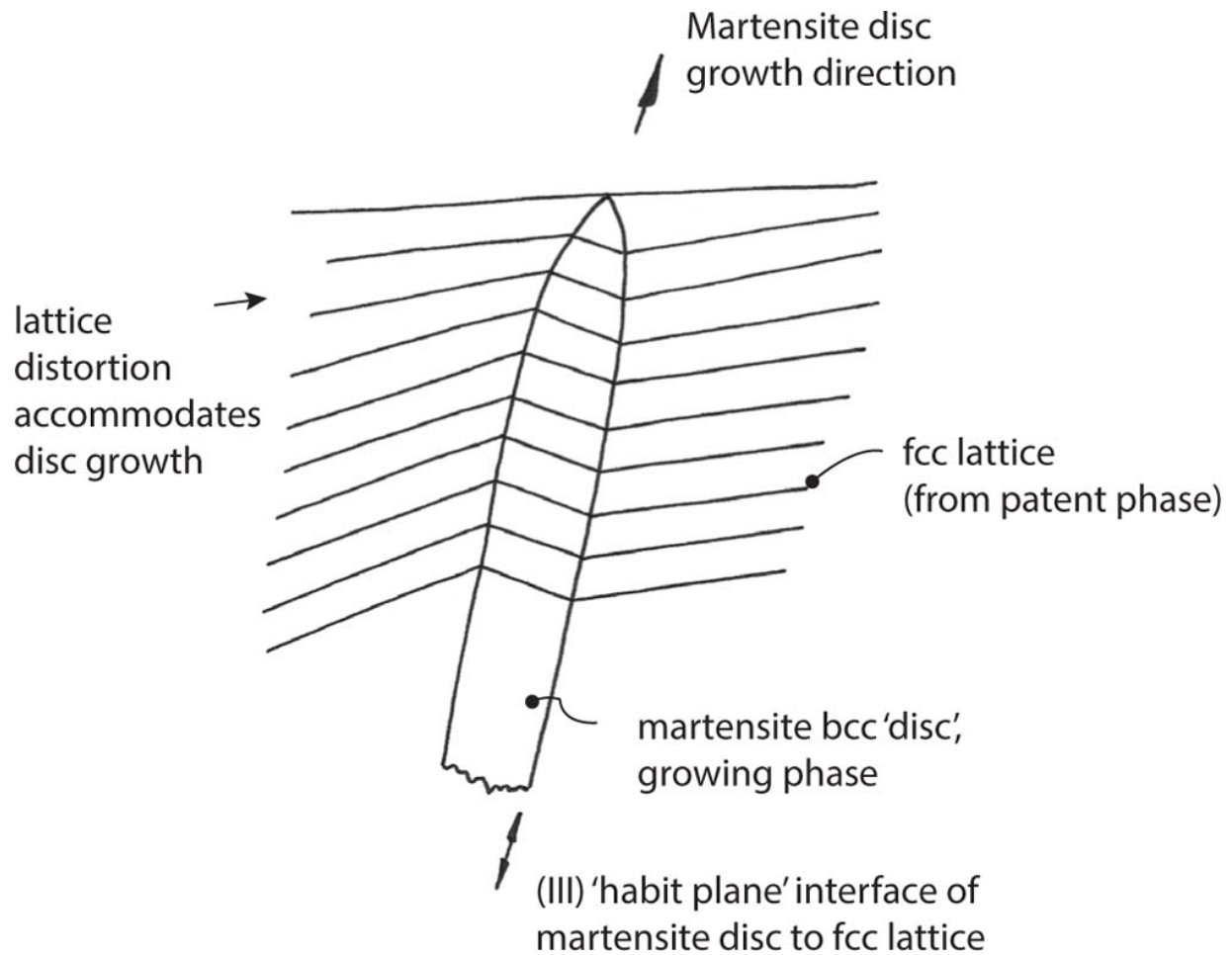


Figure 74 Martensite growth from fcc lattice, rapid switching phase.

These alloys have been used extensively on high-performance military aircraft for couplings of high pressure piping. The Nitinol fitting is retained in its expanded form at sub-zero temperatures, until it is whipped out and fitted over the pipes to be joined. On warming to room temperature, it transforms back to the high temperature phase, exerting sufficient residual stress on the tubes to effect a strong seal, capable of holding high pressure fluids. Unfortunately, whilst highly effective, the economics of this expensive alloy inhibit its more widespread application, although it can now be found in some designer spectacles which can accept a surprising degree of damage and subsequent recovery on heating, to the limit of some 10 per cent strain. Use for heat-sensitive automatic opening devices, including medical implants, is also

possible – as well as providing a simple explanation for the ‘trick’ spoon-bending antics of a certain celebrity in the recent past. Perhaps this is more a salutary tale of another ‘wonder material’, with similar martensites in the lower cost copper base alloys, which in the late twentieth century attracted considerable academic interest, research, and effort, without achieving the early promise of a universal application, which may occur in the future.

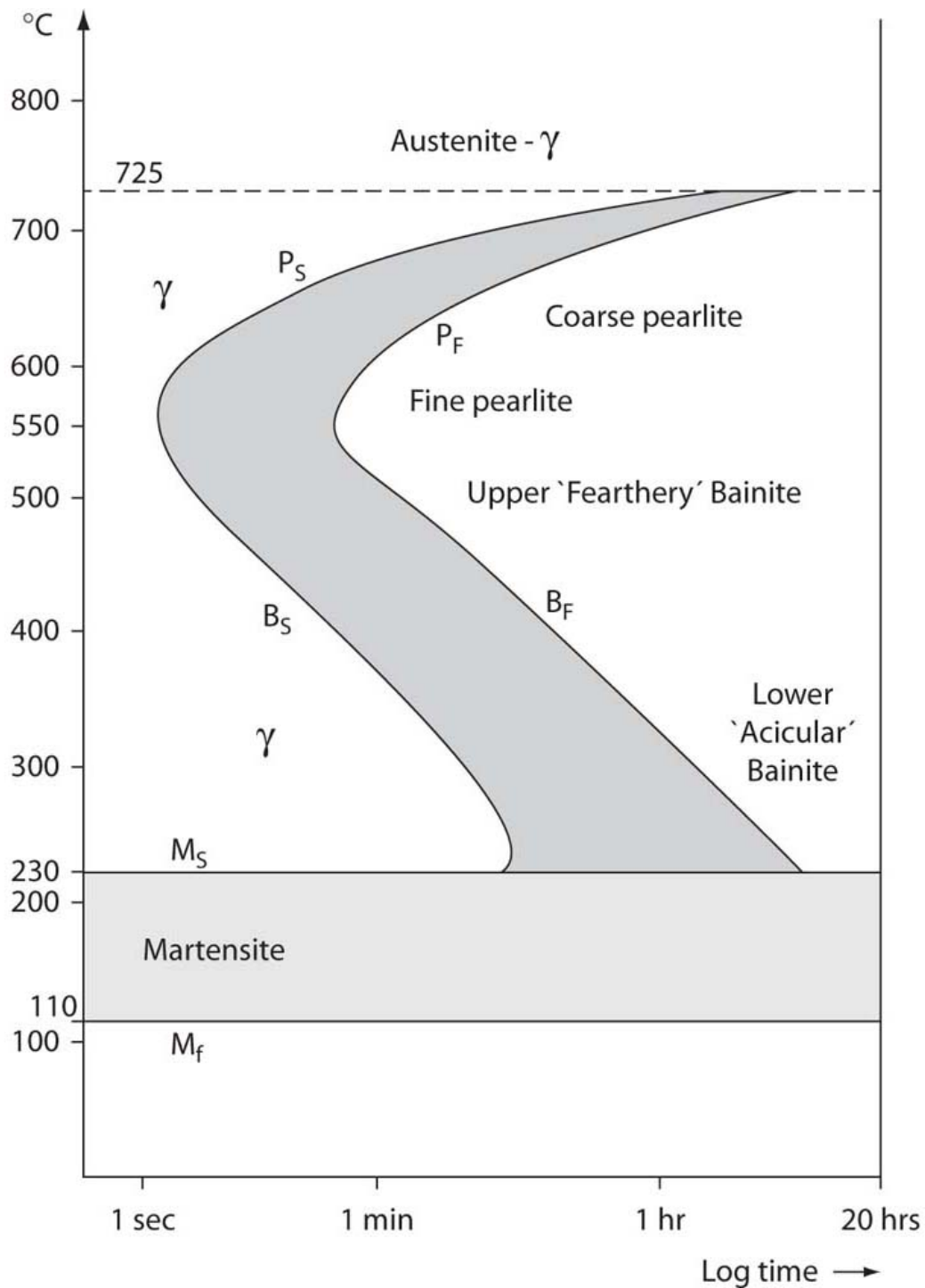
Bainite

As if martensite were not sufficiently difficult to understand, the FeC system also has a series of transformations that occur at a slower cooling rate than that required for martensite, known as the Bainitic reactions. These do not form by the same, strictly displacive, mechanism as martensite. Instead they grow, as appears under the microscope, in a more feathery but generally needle-like, somewhat less hard, structure. In steel, they form as ‘upper’ and ‘lower’ bainites, as seen below. This shows coarse and fine pearlites formed from reducing temperatures, at slower cooling rates than those for martensite. With an arrested rapid cool, it is possible to produce an upper (‘feathery’) bainite or lower (‘acicular’) bainite, according to the cooling rates described on the TTT and CCT diagrams. This is discussed later in this chapter under ‘Austempering’, another strengthening mechanism that avoids the martensite reaction, with the latter’s need for high cooling rates and consequent high hardness, and low toughness, structures. Bainitic structures can provide a highly desirable combination of strength and toughness, as an alternative route to hardening and tempering with martensite. Although the production of bainitic structures may demand relatively complex heat treatment control, it is a means widely adopted in industry for products of extremely severe and long service life, such as internal combustion engine valve springs.

As the cooling characteristics drawn on the TTT (time, temperature, transformation) diagram are often difficult to translate into practice, the CCT (continuous cooling, transformation) curves

provide a more practical guide to the behaviour of samples undergoing hardening treatments. The most significant feature is the material's section, as thickness or diameter determines the cooling rate, through the section, for any given quench medium. Whilst its skin may achieve a rapid cooling rate, the core will behave according to heat-transfer limitations, and in large items will also significantly affect the whole section's cooling rate. CCT diagrams usually provide limits to section size for the general quench mediums of water, oil or air, and the transformation products predicted for each. The limitations of the plain carbon steels in achieving full transformation across the section, even for the higher carbon types, means that only around 5/16in (8mm) diameter bar may be fully treated, even in water. This has led to the extensive development of the alloy steels.

A consequence of this difficulty in cooling at a fast rate throughout the section, in other than the specially developed alloy steels such as EN24 and 31 (AISI 4340/52100), is that different transformations can occur in practice. Typically, a fully martensitic skin, with bainitic or pearlite core, can be produced, depending on the particular circumstances of material and quench rate. This is often not a great problem and may even be used to advantage, but it is another factor that deserves consideration. In critical work there is no substitute for a sample being processed with the job, for subsequent destructive sectioning, examination by microscope, and preferably mechanical testing. The difficulty and cost of such testing may make the purchase of properly certified, fully heat-treated material a sound investment – providing it can be used 'as is'.



Note: Practically difficult to avoid 'cutting the pearlite nose' even with a rapid quench, so limited martensite formation.

Figure 75 TTT curve for Eutectoid steel (0.8%C, 0.75%Mn, eg. AISI 1080/ EN42) – typical.

Tempering

A fully martensitic high carbon steel, as produced from the heat treatment described, lacks sufficient toughness for all but the most extreme applications, where strength and hardness are needed in spite of the sacrifice of toughness, although such materials can be rendered unsafe by micro-cracking. In practice, therefore, all hardening of steel is normally accompanied by a tempering treatment to improve toughness and ductility, at the expense of some limited strength loss.

While the steel martensite brings this valuable strengthening, it is not thermally reversible, unlike that in some other systems. In the latter case re-heating simply reverses the transformation, but these martensites do not provide dramatic hardening, as indeed is the case of the maraging steels met earlier in the ultra-high performance cycle tubing. These provide a relatively ductile martensite, derived from their ultra-low carbon content.

Returning to our regular plain carbon or low alloy steels, tempering is a simple necessity for practical applications, in order to rehabilitate a martensitic structure that is potentially unstable at room temperature. This 'metastable' condition is an arrested condition, halting progress towards the equilibrium structure, although this should remain so indefinitely unless re-heated.

Tempering can be considered as a three-stage process, the lowest temperatures ($\leq 250^{\circ}\text{C}$) providing a simple relief of some of the extreme strains of the martensitic lattice, but without apparent change to the microstructure as seen in the optical microscope. The second stage is a move towards producing a more bainitic structure, a diffusion process as the steel martensite lacks reversibility, at intermediate temperatures, circa $250\text{--}450^{\circ}\text{C}$. The third stage relates to higher temperatures, towards the lower critical point (A_1), circa $450\text{--}650^{\circ}\text{C}$. This produces a significant structural change, seen in the optical microscope as a spherodization of the cementite, conferring the greatest ductility but at the expense of a further strength loss.

Practical aspects of tempering

It is traditional, for plain carbon steels, to identify the colours produced by the oxide film on the polished surface to provide a handy guide for tempering around 200–300°C. Whilst this may have served the skilled artisan well since ancient times, and provides a visually attractive prospect, I am going to abstain from its promotion as a viable method for tempering in the twenty-first century. Others may well disagree.

The first problem with using tempering colours is the difficulty in application of consistent heat during the time required which is, as before, a guide of one hour per inch (25mm) thickness, with a minimum of thirty minutes for all but very small work, where a few minutes will suffice. The easiest method for tempering is the small muffle furnace, which can be accurately set to the required temperature and held as desired, before air cooling. Care should be taken to check that the workpiece assumes close to the furnace temperature setting. This can often be easily achieved by temporary attachment of a thermocouple, providing a direct reading which can even be used as a calibration for future work, bearing in mind temperature gradients that can be significant in the larger furnaces. These simple thermocouples are now cheap and readily available from electronics retailers, in flexible wire form for use up to 350°C, and with stainless steel probes for higher temperatures. This avoids the fun and expense of sourcing and making up a thermocouple with cold junction, a method open to the dedicated purist (see [Bibliography](#) and [Appendix II: Resources](#)).

Table 38 Tempered steels – typical applications

Tempering temp. (°C)	Temper colour (°C)	Application
220	pale straw	drawing dies, shaper tools, taps
230	straw	hammer face, thread chasers, lathe tools

240	dark straw	reamers, engraving tools
260	light purple	cold chisels, axe edges
270	purple	pickaxe, press tools
280	dark purple	wood cutting tools
290	blue	screwdriver
300	dark blue	springs
330	grey	tool shanks

Alternative heating sources include oxyacetylene or oxypropane torches, but without experience it is not a simple matter to achieve an accurate temper with these. It is possible, in theory at least, to commandeer domestic sources of heat, such as the kitchen oven. The problems with heat control and temperature gradients may be solvable but it may be a slippery slope upon which one is embarked with regard to use of such domestic facilities. Perhaps the best alternative is the use of the salt bath as this is the method that was widely utilized in industry. Whilst it is still a viable option for the home worker, and the primary method advocated later for some 'special heat treatments', it is less than a simple matter to procure the requisite salts, although they are available under controlled purchase and delivery conditions that might require a group purchase. Again, it is possible to operate from first principles, for the experienced worker (see [Bibliography](#) and [Appendix II: Resources](#)). Whilst salt baths are a sound option for the experienced, a sensible approach to Health and Safety is required, as can be found from the [Bibliography](#). They do provide the prospect of accurate temperature control and, once familiar with their safe operation, efficient application.

Table 39

Material AISI 0.1 (Gauge Plate); C: 0.95, Mn: 1.2, Si: 0.25, Cr:

0.5, W: 0.5, V: 0.2		
Oil Quench 800°C, Temper at:	150°C	62 HRC
	200°C	60 HRC
	230°C	58 HRC
	290°C	56 HRC

The table above provides an indication of tempering applications, at the lower temperatures for the higher-strength applications possible with plain carbon steels, useful for experimental and comparison purposes against commercially available product, when making special tools. Particular alloys require specific data, as provided for typical ‘gauge plate’ readily available from engineering material stockists (see [Appendix II: Resources](#)).

Hardening (special processes)

The special heat treatments of interest are the well established ‘martempering’ and ‘austempering’. The former is of less interest to the home worker, deriving its benefit primarily from a method which equalizes the temperature throughout the section prior to cooling through the M_s to M_f range, and therefore implying the use of large sections. By contrast, the latter process is indeed of interest to the maker of smaller, critical parts, like springs needed on a one-off or trial basis.



Photo 131 Basic heat treatment facility for small samples, heat treatment furnace, and salt-pot heating with digital temperature monitoring. (Thermal insulation removed for clarity.)

Reference is made to the isothermal transformation diagrams, the TTT or CCT curves met earlier, which provide for the dimension of time. The lack of a time dimension on the equilibrium diagram means that it cannot provide predictions about rates of cooling required for the essential transformation products of martensite and bainites. Whilst critical cooling rate information can be found with the TTT or CCT diagrams, the former is derived from analysis of samples arrested, at various stages and cooling rates, to plot the phases at obtained at each point. These are the isothermal (time–temperature–transformation) graphs, essential to understanding and prediction of the phase and consequently mechanical properties of steels. The Continuous Cooling-Transformation plots deviate a little from the TTT curves, and provide a more practical basis for the prediction of properties, but for general purposes they can be considered together.

The martempering operation can now be explained from [Figure 76](#), which illustrates conventional hardening on a TTT curve, displaying a cooling rate that should result in producing the required phases. The martempering curve takes the greater time to reach ‘pearlite nose’ on the TTT diagram, a characteristic of the low alloy

steels, in order to cool rapidly enough to miss the onset of pearlite at higher temperatures, then halting prior to the M_s . The delay allows the work to achieve a constant temperature throughout the section, and is then cooled through the M_s to M_f range, thereby imparting less thermal shock and producing a more uniform transformation. This is beneficial in avoiding quench cracking and distortion, but is only applicable given the favourable TTT profile of low alloy steels.

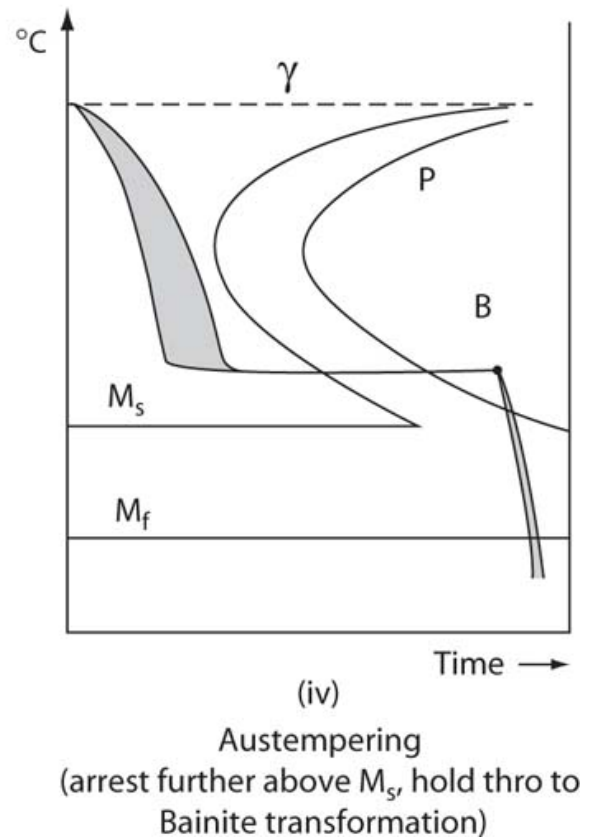
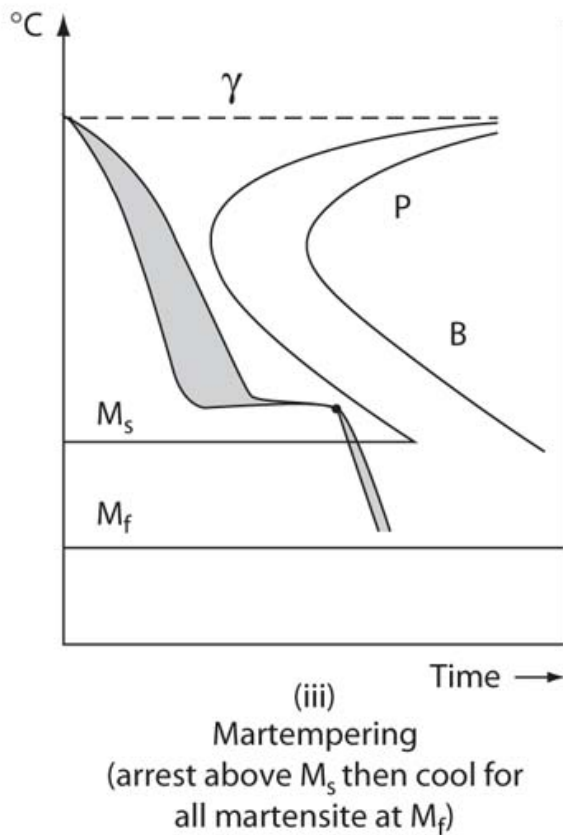
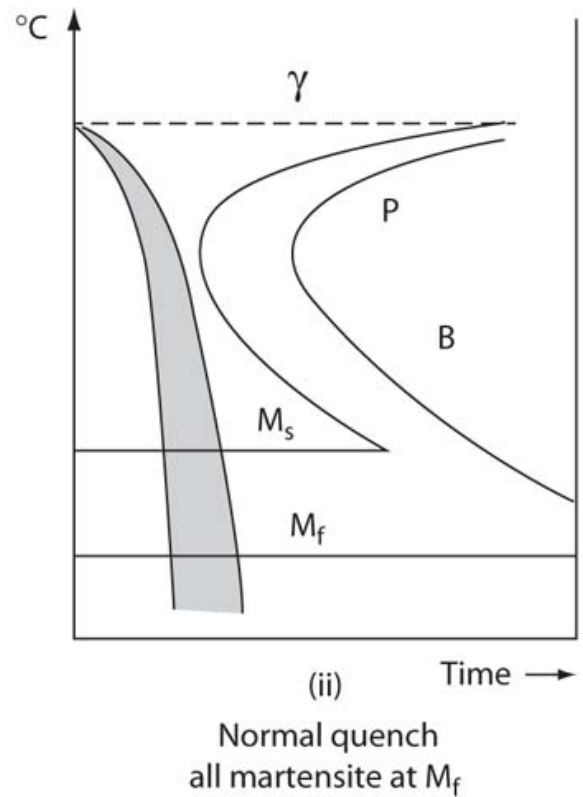
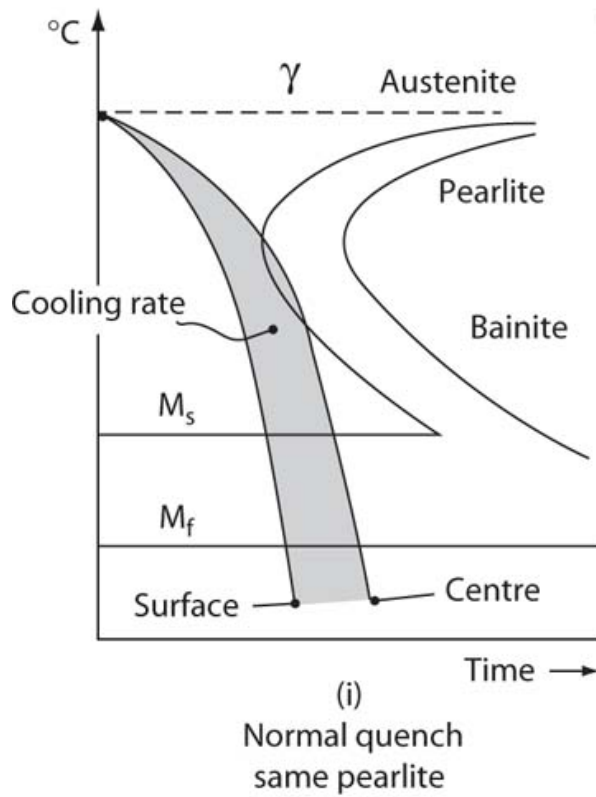


Figure 76 Normal quench, Martempering and Austempering TTT diagrams.

Table 40 For a 0.95% carbon steel (similar to silver steel).

Heat Treatment	(H _{RC}) Hardness	(kg/cm) Impact Strength	Elongation
Water Qu & T	53	180	1
Martempered & T	53	330	2
Austempered	52	590	10

The austempering process, as seen in [Figure 76](#), adopts a similar approach to the interrupted cooling method of martempering, but this time the isothermal hold is maintained above the M_s and right into the lower bainite section. This produces a fully bainitic structure, instead of martensite, which has highly significant benefits for toughness, without loss of strength, as demonstrated in [Table 40](#).

This process is particularly beneficial for highly stressed steels requiring a high toughness and lack of crack sensitivity, which may not be available by conventional quench and temper methods. Springs, and rocker arms for internal combustion engines, are prime examples. The downside is the relative difficulty in effecting such treatments, the most practical route being to quench from the muffle furnace into a salt bath at around 300–400°C, with appropriate attention to Health and Safety, to hold until cooling in air to room temperature. Holding times for small items are typically only a few hours, as shown in [Figure 78](#). This is a process most suited to low alloy steels, or cast irons, where the cooling rate enables the pearlite ‘nose’ to be avoided, and the structure can be fully transformed into a bainitic one. Thus, in [Table 40](#), this can only be achieved in small pieces, with rapid cooling rates.

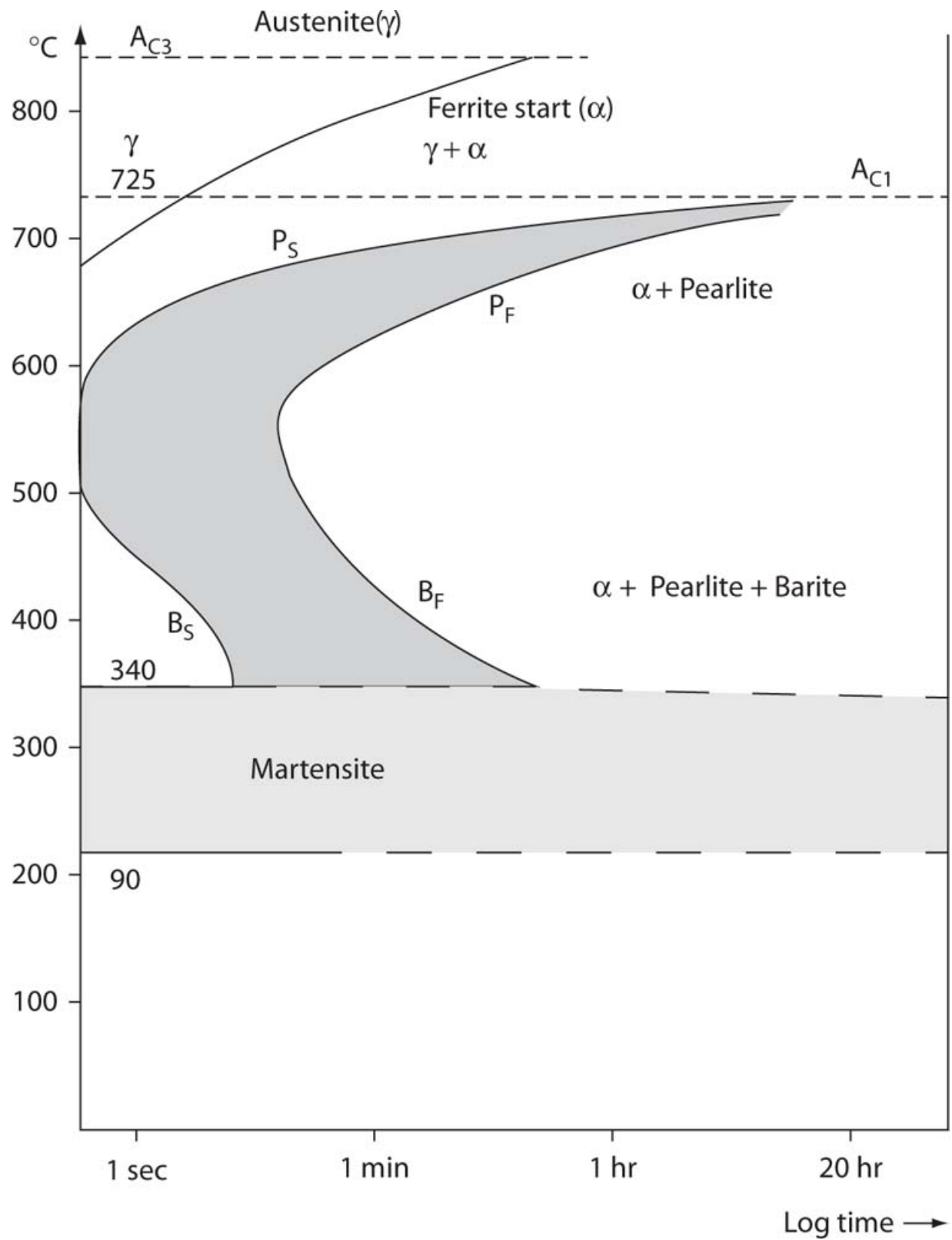


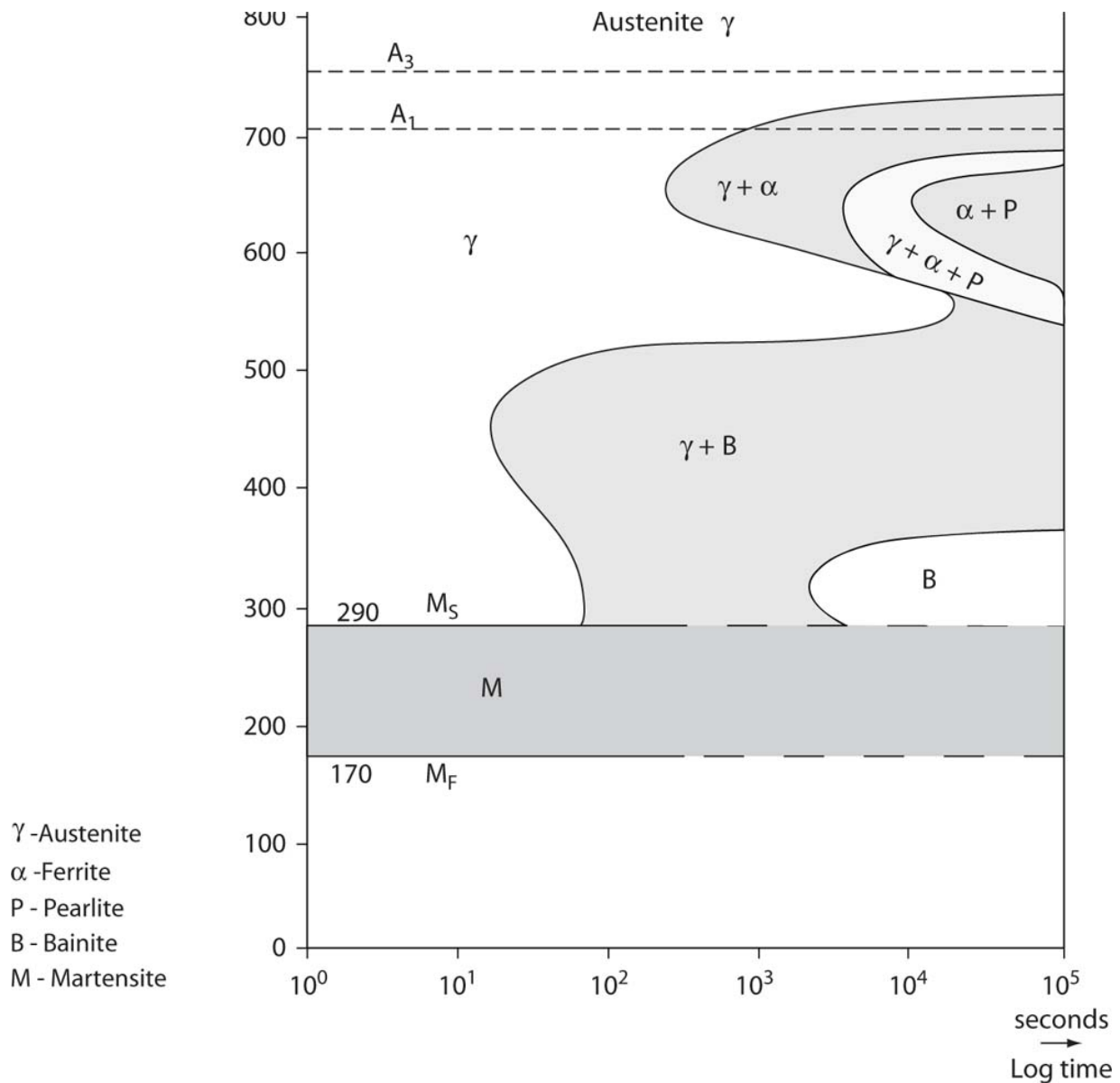
Figure 77 TTT curve for 0.3%C steel (eg. AISI 1030/EN5).

A final process is 'ausforming', which involves extensive working of the steel at a temperature a little below the re-crystallization temperature, thus building up a high degree of dislocation density. For a suitable low alloy steel TTT diagram, having a relatively long time whilst above the M_s to reach the bainite start (B_s), this is essentially an adaptation of the martempering process with a high level of deformation, up to 90 per cent reduction, applied prior to cooling through the M_s to M_f range. This is the process applied to the maraging steels, met earlier in ultra-high-strength tubes, for cycle frames and top-end sports equipment.

Alloy steels and transformation diagrams

The low alloy steels introduce small additions of alloying elements such as Cr, Ni, and Mo, to the FeC diagram, in order to modify the TTT curve so that the extremely rapid cooling rates, required to harden the plain C steels, can be relaxed. This produces a far more manageable heat treatment process, and can produce the desired transformations in larger sections, with greatly improved through-thickness properties. The extent of hardening, and comparative hardenability, has led to various standardized methods, such as the 'Jominy End Quench' test (refer to the [Bibliography](#) for details).

Examples of TTT diagrams for carbon and low alloy steels are shown in [Figures 77](#) and [78](#).



Note: Displacement & pearlite nose enables practical cooling rate for full transformation to martensite, or for martempering or austempering

Figure 78 TTT curve for low-alloy steel, eg. AISI 4340/EN24 (typical).

Quenching

In order to achieve the desired quench rate the general choice is between water, oil and air as the medium. Recent developments include a range of polymer quenchants, but these don't seem to be readily available to the amateur. We are therefore forced to bulk

purchase a commercial grade through a group, or experiment with others, observing Health and Safety, such as for flash point and off-gassing. The use of water is safe, given simple precautions in handling very hot material, and avoidance of burns or excessive steam generation by using a sensibly sized container. The process of water quenching needs to accommodate the heat transfer characteristics described as the three-stage process. This involves an initial vapour blanket effect at high temperature; followed by a vapour boiling stage; and finally a liquid cooling stage to room temperature.

The method of quenching requires some thought and practice, as a simple static immersion, or frenetic splashing about of the workpiece, will maximize the above stages. The best technique is a smooth, rapid immersion into the quenchant, in as short a route from furnace as possible. Once immersed, a steady, not frantic, agitation preferably with a spiral motion, is used to disperse the near-work blanket that forms in the liquid. For the limited sizes of pieces likely to be handled, the time involved is short. The temperature range is defined by the transformation diagram, and good practice is to retain the piece held in tongs until fully cooled, before withdrawal.



Photo 132 Direct hardening of steel bar into carburizing (case hardening) compound.

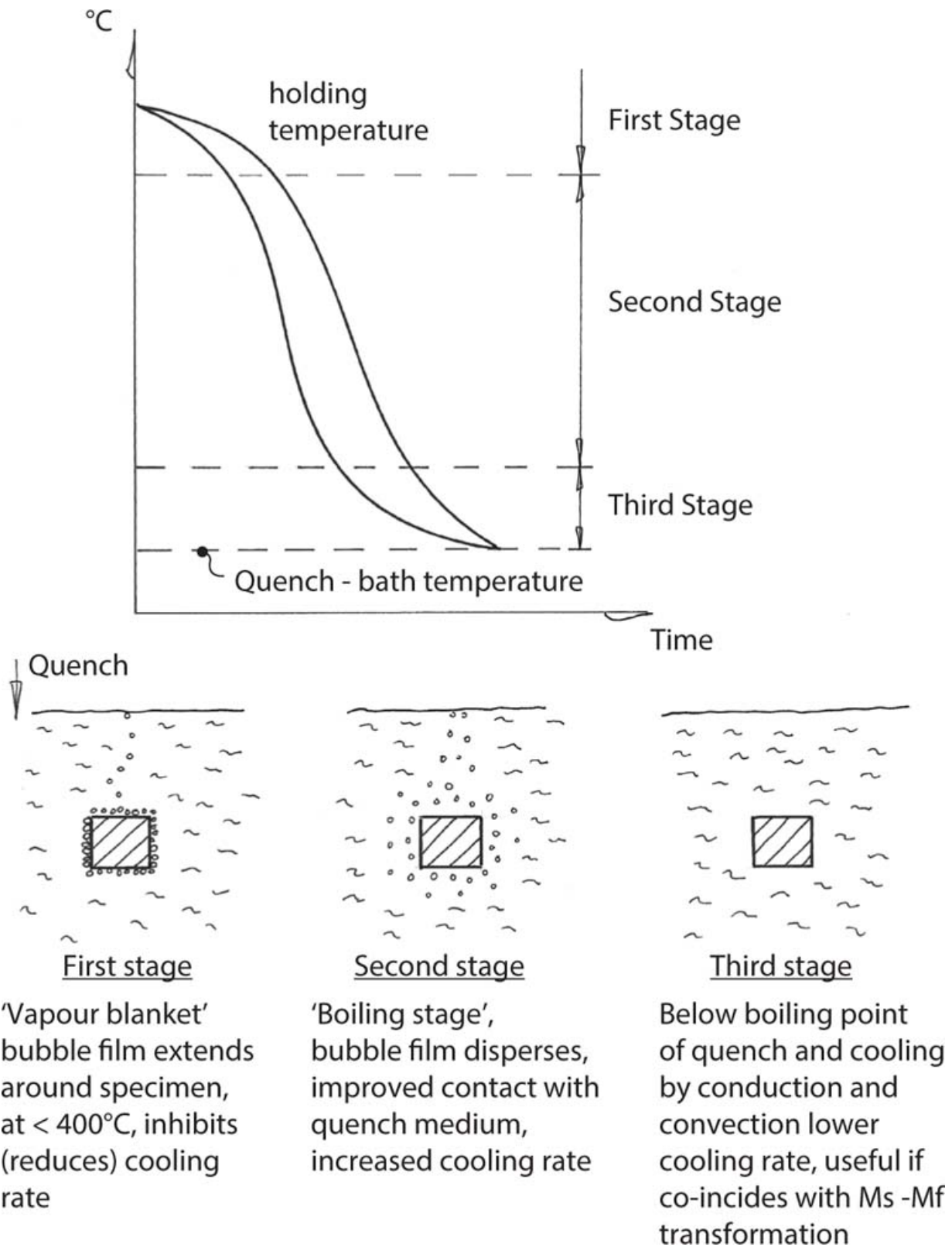


Figure 79 Typical quenching characteristics, e.g. water.

Case hardening

Case hardening, the familiar form of 'carburizing', is just one method from a multitude of surface-hardening processes, from flame and induction hardening, to surface treatments like Nitriding and many similar commercial processes. However, for the home workshop it is case hardening that is the most attractive, due to its practical simplicity, effective results, and low cost. All of the other processes, whilst highly effective, are not appropriate for the small scale operation, not least on the grounds of cost. There is a very good case to be made for this process as many items require a hard surface, for the purposes of wear resistance or facility as a bearing, and to inhibit microcracks that can be the pre-cursor of fatigue. Often the hardness is only required at the surface, and a core with a greater toughness may be preferred.

Care must be exercised in a design where the attractions of manufacturing, say, a crankpin from mild steel with a case-hardened surface, may obscure a need for a higher shear strength than the mild steel can provide. What is perfectly appropriate for a model steam engine crankpin may be unsuitable for a full-scale high compression internal combustion engine crankpin, which could lead to fatigue failure in service. Similarly, items subject to occasional overload conditions in safety-critical components, like cycle and motor cycle wheel spindles, require through thickness properties that case hardening will not provide. Nevertheless, especially on models, case hardening provides an excellent solution to sliding and wearing parts, on the principle of one hard part mating with a soft bearing having already been established in [Chapter 4](#).

The theory of case hardening has already been established from the hardening of carbon steels, as it involves taking a low carbon steel (typical mild steel <0.2 per cent C) and diffusing an excess of carbon atoms into the surface, often to a depth of only a few thou (2–20 thou, 0.05–0.5mm) being sufficient. There are two principal methods for practical case hardening: the first is packing a container ('box-carburizing') with case hardening compound to envelope the

workpiece, fitting a lid and subjecting the item to the time and temperature in the furnace as directed by the compound's suppliers. This requires a muffle furnace and is a treatment usually lasting several hours for a reasonable case depth. The other method, often known as 'hearth carburizing', involves simply dipping the work, heated to around 800°C, cherry red, into a container of sufficient case hardening compound to cover it, withdrawing once the compound has picked up on the piece, applying some more heat to fuse the granules, then allowing to cool in air, cleaning off the surplus and then repeating the process according to the depth of case required, each treatment effecting a case depth of around one or two thou (0.02–0.05mm). Note that for the most efficient absorption of carbon, heating above A3 is preferred, at around 920°C in mild steel.

A particular advantage of the process is that, whilst the surface may have achieved a depth of carbon equivalent to a thin layer of medium carbon steel, it is not in the hardened, martensitic, state yet. This means that it can still be readily machined, or worked with hand tools in areas where the hardened surface is not required, such as the threaded ends of a shaft where only the bearing surface is required to be hard. The other parts can be turned and screw-cut at this stage, removing the case. When ready, the section to be hardened is then subject to a conventional heat treatment (heat above upper critical line and water quench), which will render the required section as hard as an equivalent fully hardened carbon steel, whilst the other parts will be unaffected. The only difficulty may arise from the surface finish of the portion requiring heat treatment, but this may be preserved with a suitable surface protecting wash or, provided there is sufficient case depth, a light grinding operation to finished size.

METALLURGICAL ASPECTS OF JOINING: FUSION AND NON-FUSION

In this concluding section we consider, very briefly, some of the aspects of joining by fusion welding, as this is such an important feature in the manufacture of product large or small, and is commonly regarded as something of a 'black art' by otherwise well-informed engineers, amateur and professional alike.

Without doubt, the scope of the subject is enormous, but from the topics discussed so far, it is possible to gain a rational understanding of the processes of fusion welding in different metal systems; those which it is sensible to pursue, and those not to touch with a bargepole, or at least welding electrode! Fortunately, the good old low carbon, mild steels are very tolerant of such joining, and are thoroughly explored in the literature, at least with regard to the different welding processes, welding consumable handling, and manual techniques (see [Bibliography](#)).

That attention is not normally directed to the fundamental metallurgical aspects of the process does not normally cause practical problems with mild steel, beyond the [introduction](#) of defects attributable to poor welder technique, joint arrangement, or preparation. These steels generally permit substantial use and abuse whilst still delivering joints of sufficient strength and integrity despite some faults. But where applications are safety-critical, it becomes appropriate to rely on a more rational approach to design of joints, avoidance of defects, and testing; rather than 'chancing it' in ignorance of pitfalls. As most engineering failures derive from fatigue, so do most fatigue failures in fabrications originate from welded joints. Whilst it is not at all difficult to produce a butt weld in mild steel with around 100 per cent efficiency (no loss of strength through the joint), it is also equally possible to produce a visually sound fillet weld with sufficient lack of fusion as to be worse than useless. The watchword is to proceed with caution and gain experience steadily, with a practical grasp of materials, joint design, and defects, before tackling critical work.



Photo 133 Re-heating test bar after dipping in case hardening compound.



Photo 134 Heat treatment sample, ready for removal from furnace after soaking at temperature.



Photo 135 Case hardening sample, stage I, machined oversize, mild steel in as-received condition.

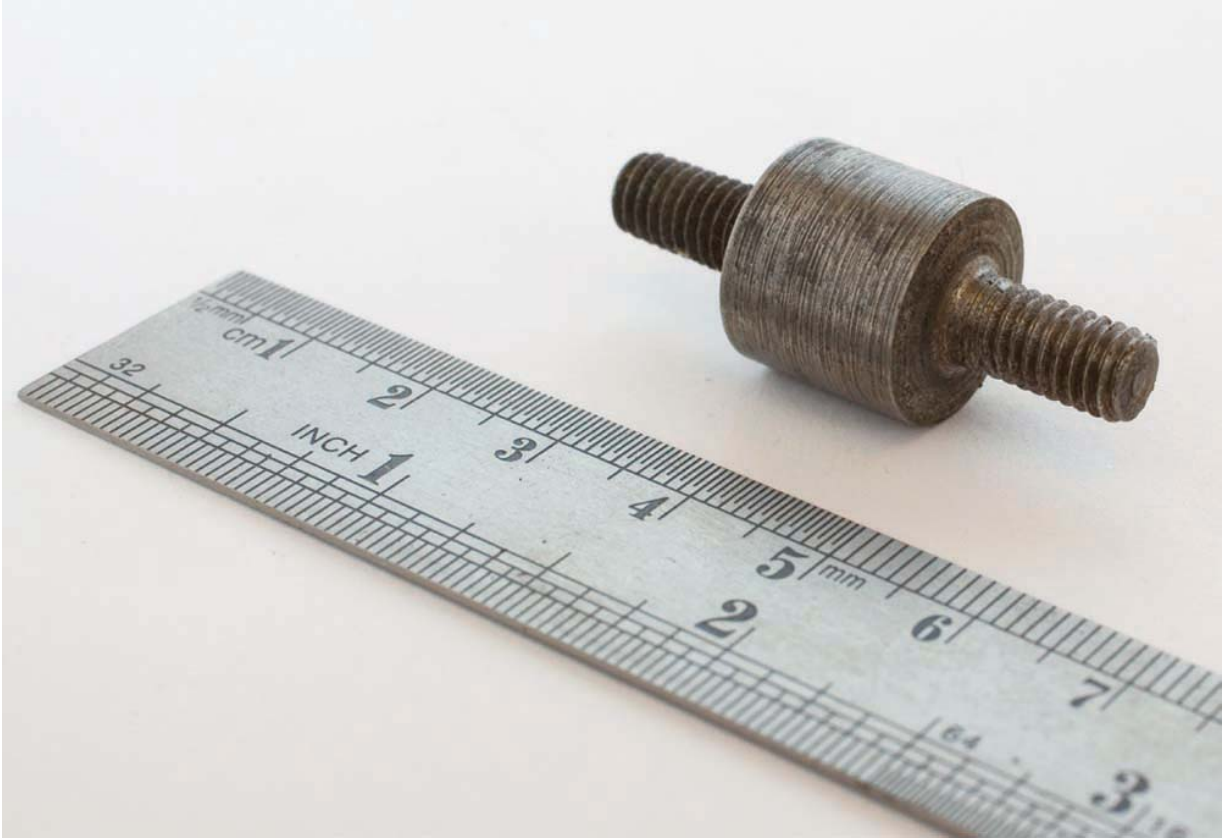
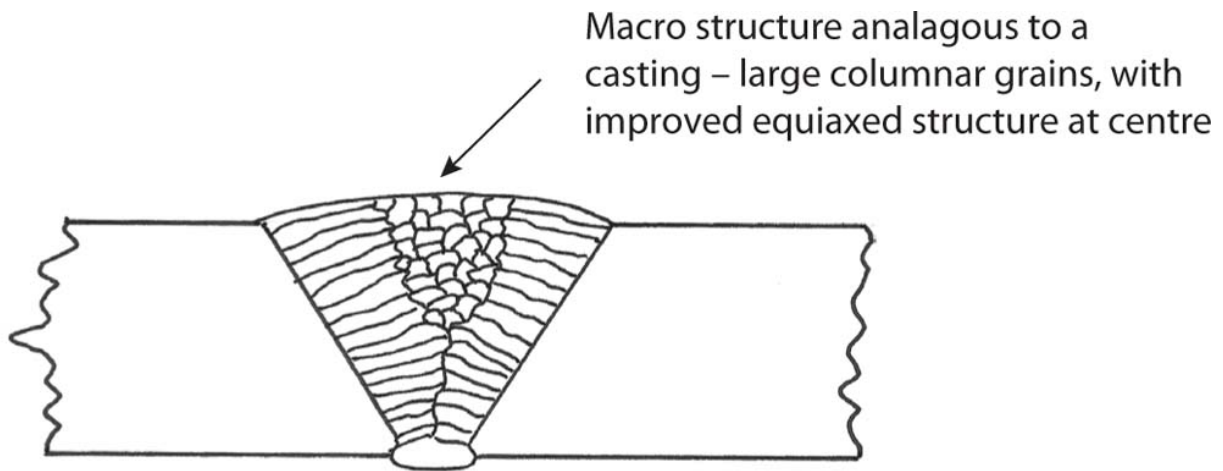


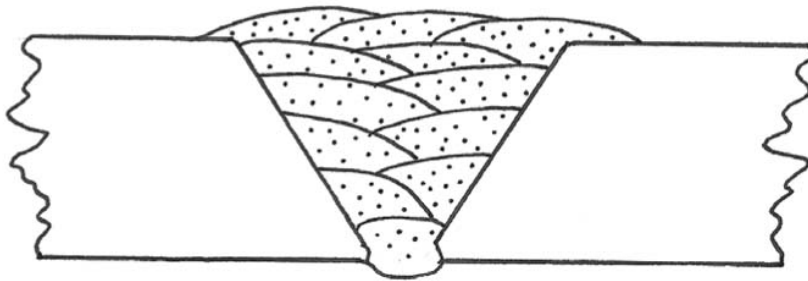
Photo 136 Case hardening sample, case treated, protected threads.



Photo 137 Case hardening completed by quenching and clean-up. Plain surface hard, threads as received.



- (i) Single run, large plate,
(typically, heavy deposit from submerged Arc
welding, SAW)



- (ii) Multi-run, small beads, each bead improves previous. Final,
'capping' runs provide 'temper beads' to improve adjacent
structure, especially for higher carbon content/
'carbon equivalent', (CE)

Figure 80 Welding structures typical for heavy single-run and small-bead multi-run welds.

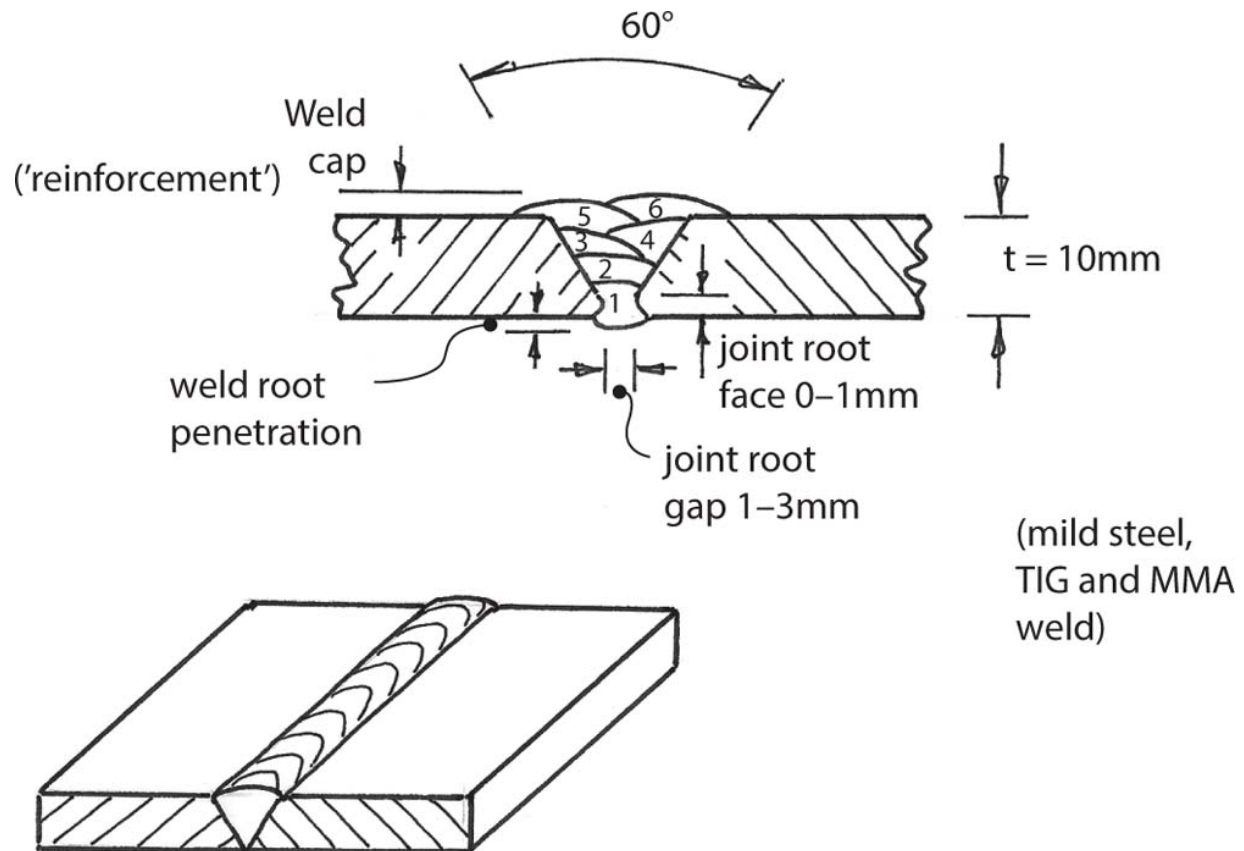


Figure 81 Butt weld profile – typical.

The weld

As mentioned in the earlier section on castings, the normal process of arc welding provides, in essence, a miniature arc furnace, made viable by either a gas shield, as for MIG (manual inert gas) and TIG (tungsten inert gas), or flux deposit, for MMA (manual metal arc) or SAW (submerged arc weld) processes, which protects the transfer of molten metal across the arc from consumable to workpiece joint. The temperatures in the plasma of the arc are in the order of 5000°C and the consumable, filler material, is transferred as molten metal, arriving with sufficient energy to effect full fusion, either with the joint wall or existing deposited weld. The weldment then cools and solidifies, in the manner of a miniature casting, as the molten weld pool continues along its route across the joint.

This brings with it the solidification processes akin to the larger scale casting, with the features discussed in that section, but a well

made weldment will normally comprise a number of small weld beads, limiting the deficiencies of the start of solidification, and avoiding the large columnar grain growth at the larger scale. For the processes employed in the home workshop, we need not worry about the problems in deposition of large weld beads, in deep joints, with coarse weld structures. The typical weld bead for small equipment is only around 3mm diameter or less, so that filling a 10mm butt joint can be achieved with a surprisingly larger number of runs. Although appearing a time consuming process, production rate is not normally at a premium, as it is for the professional. Small beads tend towards optimum properties, in an analogous manner to the small grains, which provide the best strength and toughness route (as discussed earlier). Although even the smallest weld bead is comprised of many grains, the effect on dislocation flow, and therefore yield strength, is similar. Thus, a properly made weld has little difficulty in matching or exceeding the parent metal strength, given the appropriate consumable and process choice.

The HAZ

Having established that, in theory and good practice at least, the butt weld provides no loss in strength, the real difficulty lies in its effect on the surrounding material. This is the region from the welding fusion line, into the parent metal; a few mm, normally up to 5mm, it is known as the 'heat affected zone' (HAZ). Once again, mild steels are normally able to accommodate the changes which occur in the HAZ without significant problems, as this region is one that draws on the understanding of the effect of cooling rates from high temperature, which is basically as discussed in the section on 'heat treatment of steels'. The understanding of HAZ reactions draws on the behaviour of the parent metal from the TTT diagram, although the situation is complicated by the fact that there is a severe temperature gradient, from the fusion line at circa 1500°C, to room temperature (or pre-heat, or weld 'interpass' temperature), which exists at the parent plate. As this gradient is steep, the critical

range in steels is taken conventionally as 800–500°C, and the extent of the HAZ is <5mm. However, there are several regions within the HAZ, produced as a consequence of their different cooling rates. In the mild steels, the most critical region is normally that very close (<1mm) to the fusion line where, as remarked earlier, it is possible to achieve hardnesses significantly greater than would be expected in conventional heat treatment. In practice this can normally be neglected.

Carbon and alloy steels

The situation regarding welding of the medium to high-carbon steels, or low-alloy steels, demands a proper understanding and knowledge of the alloy before attempting operations. For reclaimed material there is often a problem in identification, without which welding should be regarded as non-viable, bearing in mind where applications of such alloys can be expected to experience high-stress service. Given sufficient data on the steel, it could be perfectly sensible to undertake welding, but only with special controls and checks to ensure that the HAZ hardness remains within required limits. As the amounts of carbon and alloying elements increase, so does the difficulty in effecting a safe weld. The use of simple formulae for 'carbon equivalent' (CE) provides a guide to the increasing hardenability, and loss of toughness, that goes with increasing CE value.

Table 41 CE – carbon equivalent for welding (short formula, commonly used, from the IIW – International Institute of Welding).

CE = C + (Mn) / 6 + (Cr + Mo + V) / 5 + (Ni + Cu) / 15	
CE	minimum weld pre-heat (°C)
< 0.45	< 100
0.45 – 0.60	100 - 250

> 0.60	250 – 450
CE and preheat guide for plain and low alloy steels; but pre-heat and weldability also strongly depends on several other factors, including section size, heat input, joint restraint, contamination, welding process and welding consumable. These conditions can render even mild steel of restricted weldability.	

In order to handle anything above the CE of a mild steel, CE ~0.43, such as EN8 or EN19, it is essential to control the HAZ cooling rate. The only means of achieving this is through the application of 'pre-heat', and the juggling of the electrical heat input. The latter is a function of electrical power (watts = volts × amps); weld pool travel speed; and process efficiency, where:

$$\text{heat input (E} \times 10^{-3} \text{ kJ/mm)} = \text{process efficiency (} \sim 0.7 \text{ to } 0.9) \times V \times l/\text{travel speed (mm/s)}.$$

As would be expected from the TTT diagram, the rapid cooling rates in the HAZ will produce a range of transformation products, including martensite. This can produce a sufficiently high hardness to reduce the toughness to dangerous levels, making the HAZ highly crack-sensitive. For the low alloy steels, such as EN19/24, AISI 4140/4340, it is usually necessary to apply a pre-heat of some 300°C, and retain the job above this temperature pre- and post-welding. The job is thereby kept above the M_s , before transferring directly to a post-weld heat treatment (PWHT) furnace for stress relieving. If desired, a subsequent heat treatment could be undertaken, but usually such products are used as-welded, and it would be sensible to undertake the testing described earlier, on a disposable sample.

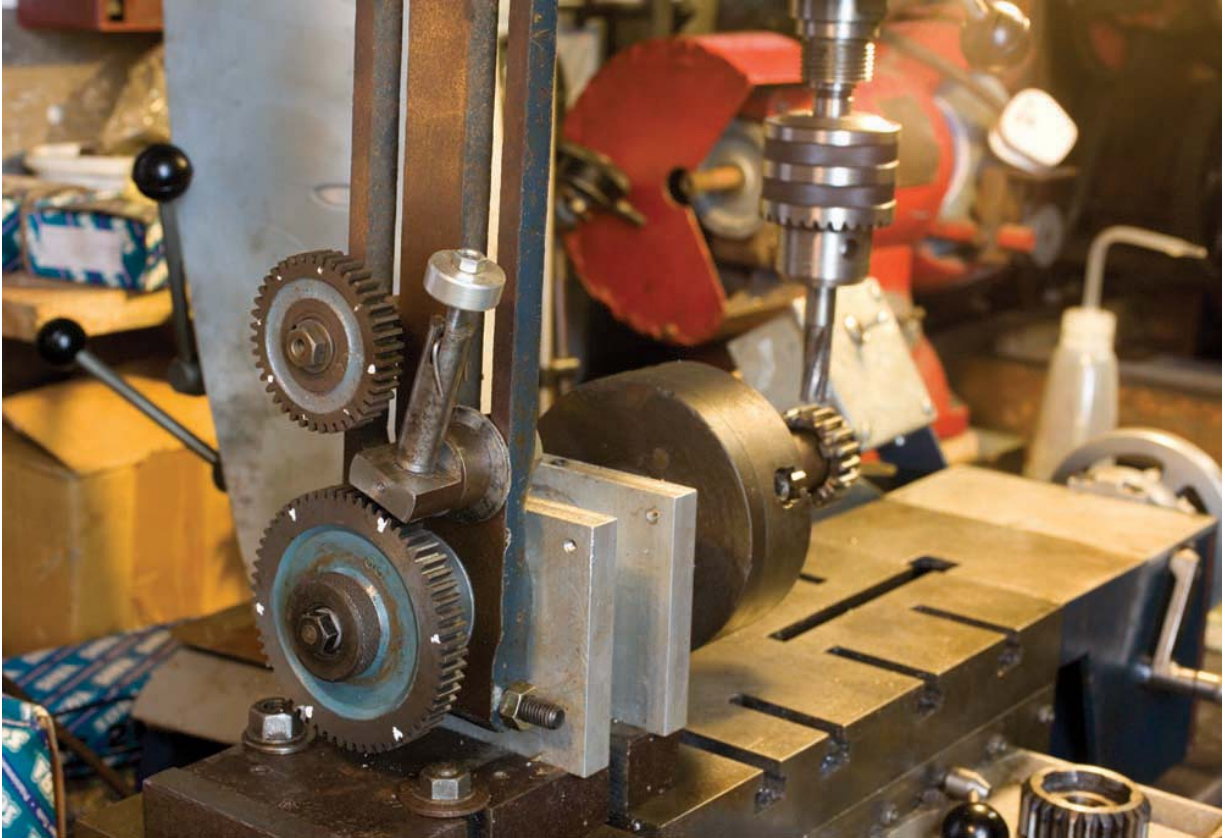


Photo 138 Worn gear dogs, faces rebuilt with alloy 625 weld, ready for indexed machining.

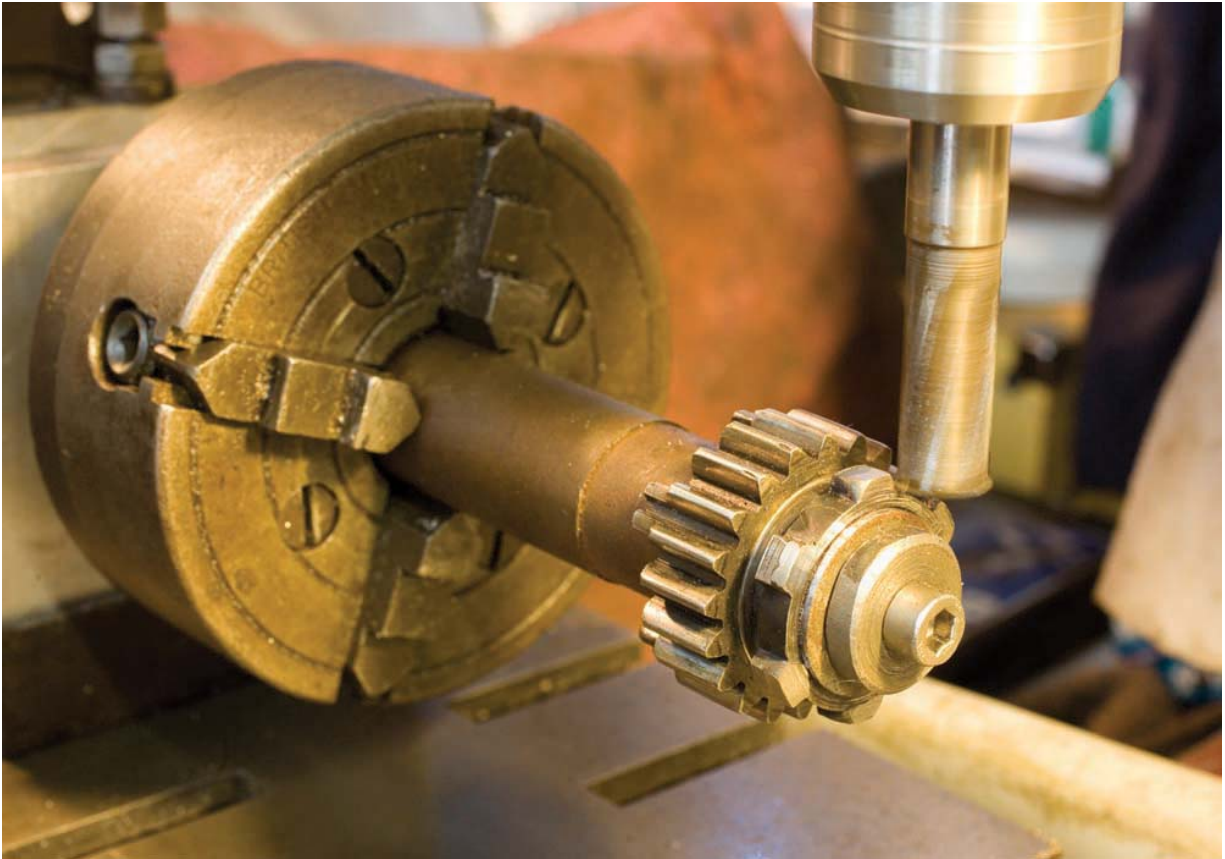


Photo 139 Machining of repaired gear in progress.

Weld stresses

A vital feature, in the performance of any welded joint, is distortion and residual stress, also discussed [Chapter 3](#), that accompanies every weldment. The stresses created by a weld are highly significant, and occur due to shrinkage in both the transverse and longitudinal directions, in the ratio of approximately 10:1, respectively. Therefore, a butt welded, thin-walled tube, say 1mm wall, can rather surprisingly undergo a significant dip and pinching due to the transverse and circumferential shrinkage, where the parent material has insufficient strength to resist the solidification stresses of the weld. For tubes of greater strength, say 10mm thick, these strains are accommodated internally as residual stress, but it is the case that the transverse shrinkage causes the local parent metal to be strained past yield stress. With ductile materials like mild

steel this is not normally a problem, but even here it is possible to produce a design with sufficient joint restraint to make it certain that the weldment will suffer failure on solidification!

This effect is clearly seen when welding plates, either flat as butt welds, or as a 'Tee' shape with fillet weld on one side only. If unrestrained, these will deform, unless of massive section. The solution is to restrain prior to welding, often by welding on 'strongback' restraining plates, as the forces exceed yield in the adjacent parent material. After release of the restraint, there is no plastic movement only the, normally insignificant, element of *elastic* spring-back. However, the residual stresses remain hidden within the material, and where this could be a problem, such as in a fatigue environment, a PWHT operation, holding at 600°C for steels, is required.

A number of techniques can be used to mitigate the effects of distortion, including balancing weld runs by alternating welding on either side of a butt joint, or double-sided fillet weld. But note that the residual stress still remains; it is just that it is more balanced and the effects are internalized. In many jobs, the control of distortion is the limiting factor in a successful outcome, and distortion due to welding is the enemy of welding repairs to over-machined work. This can often limit the otherwise attractive prospect of repairing by building up a worn shaft with welding. Such repairs need to be balanced, but may still provide problems on re-machining with release of residual stresses causing unwanted distortion. Experience is the only means to determine the limits of such work, and correction by *ad hoc* methods such as spot heating may be needed as a last resort.

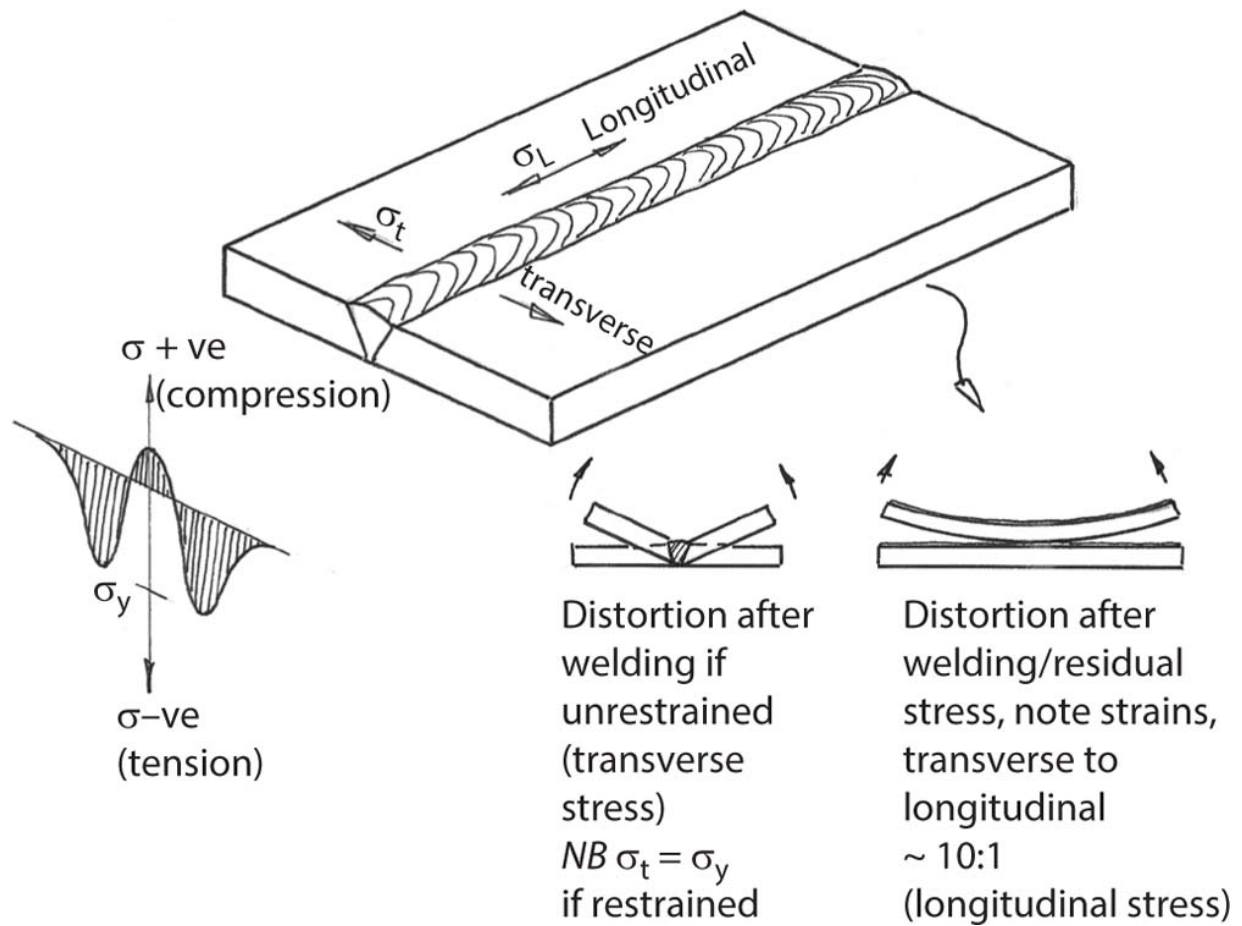


Figure 82 Welding residual stresses.

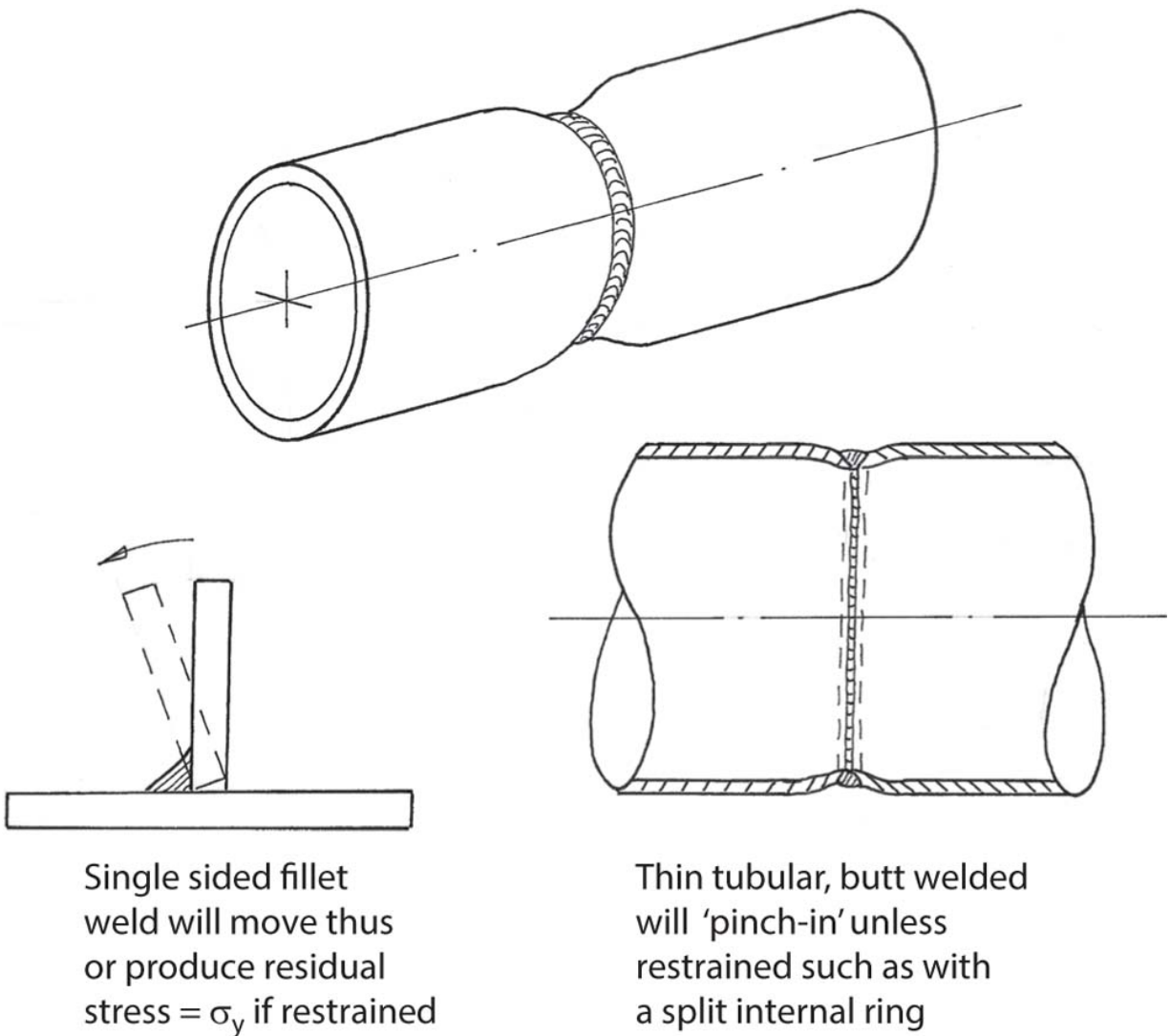


Figure 83 Weld distortion tendency in thin-wall tubing.

Although steel is the material most likely to be welded in the home workshop, it is possible to weld some of the light alloys, but only the non-heat treatable 1000- and 5000-series, or the 6000-series with subsequent heat treatment to restore full strength. Oxyacetylene welding of the aluminiums is a demanding technique, due to the lack of surface indications before reaching the liquid state, and the difficulty of dissipating the tenacious oxide film. TIG welding is preferred, but requires expensive equipment, with AC (alternating current) necessary to disperse the oxide film. However, given these facilities, the light alloys present relatively few metallurgical

problems, compared to the steels, although distortion control requires equal care in joint design and general fabrication planning.

Non-fusion processes

Finally, there is a considerable array of non-fusion approaches to joining, based on diffusion mechanisms and requiring substantially lower temperatures. Particularly of interest for the home workshop is brazing or bronze welding, which utilizes a copper-base alloy consumable. This is melted onto a base metal, heated well below its melting point, but sufficiently hot to enable the rapid diffusion of the consumable into its surface layers. Thus, the steels, and even some cast irons, can be bronze welded at $\sim 800^{\circ}\text{C}$. This facility has been used to avoid the complex problems in the fusion welded HAZ of some steels, and was the method almost exclusively used in high performance tubing such as Renolds 531, widely employed in high-duty competition motorcycle frames from the 1930s to 1990s. However, the ultimate properties of this and more advanced steels are now better approached with properly controlled fusion welding, normally by the TIG process.



Photo 140 Traditional 'tools of the trade'.

In conclusion, the well established silver soldering process is still pre-eminent in the manufacture of the copper boilers that form the heart of the model steam loco, traction engine, or stationary steam engine, as has been discussed in [Case Study 1](#) of [Chapter 1](#), and completes this brief review of engineering materials, with the [Bibliography](#) alluding to the pursuit of further information.

Appendix I: Hardness measurement

BACKGROUND

Mechanical testing of metals provides the basis for all analysis of the strength of materials, but as explained in [Chapter 3](#) it requires the use of specialist equipment normally only available in test laboratories. We are particularly interested in the characteristics of the stress–strain graph, but the one method of getting a reasonable idea of strength, without the cost of full tensile testing, is through the hardness test.

Hardness testing provides a superficially simple and accessible method which gives at least an approximate measure of strength. By using charts for steel (see [Table 42](#)), it is possible to assess the approximate condition and effects of heat treatment, as described in [Chapter 5](#).

The measurement of hardness is apparently simple, involving either the small indentation of the surface by a ball or pyramid indenter, or the firing of an indenter at the surface. In the former, the indent is measured, and in the latter it is the ‘bounce’ height or velocity that provides a measure of hardness. However, a thorough analysis of the process is rather too involved to be explored fully here, and the simple testing described below satisfies our limited requirement for characterization, particularly with regard to the implied strength of the hardenable alloys of steel and aluminium.

The most popular measures of hardness are those shown in [Table 42](#). These scales are useful because of the large body of experience that has been established, analogous to Charpy impact test data. Probably the method of most universal usage is the ‘Vickers Hardness’ which uses a pyramidal diamond indent, which can be

measured across each diagonal, then related to Vickers Hardness (H_V). The geometrical indent lends itself to different loads, with 30kg providing a relatively large indent, and therefore covering a larger part of the microstructure than the smaller loadings such as 10kg. A 'micro-hardness' of 2kg and below provides a small-scale indent that measures a small region, such as the variations across a weld HAZ where displacements of 0.1mm can be critical.

The traditional Vickers machine, with manual controls (see [Photos 93–95](#)), is still used in laboratories as a standard, but smaller electronic variants can now be used for in-situ work. These hand-held devices are usually based on the 'Loeb' principle, which measures the velocity of a small indenter on the rebound from the surface. Whilst convenient, application can be sometimes subject to difficulty in interpretation, as scatter in results of hardness testing is normal and experience is required in all applications.

The other methods most often employed are the Brinell Hardness (H_B) and Rockwell Hardness (H_R). They are similar in general principle to the Vickers test, the former using a ball indenter, with larger weights, intended to produce a larger indent and thus a more general measure of hardness that is less sensitive to local high hardness, which can be detected by a micro-hardness test. The latter obtains a hardness by measuring the indent load, rather than the size of the indent as for Brinell and Vickers.

The Rockwell tester, usually the 'C' scale, as H_{RC} , also uses a hardened ball indenter, designed to measure the upper range of hardnesses, on a larger indent than the Vickers. It is used particularly for high-strength, low-alloy steels and is often quoted for heat-treated silver steel and gauge plate.

Photos 141 and 142 show portable hardness testers operating on a mechanical principle, with a conical indenter and readout on Vickers, Brinell and Rockwell scales. As with the portable electronic versions, these can provide a useful estimate of hardness, but for more accurate work it is necessary to move into the laboratory for dedicated equipment, as seen in [Chapter 5](#).



Photo 141 Hand-held mechanical hardness tester with multi-scale reading, for approximate hardness measurement.



Photo 142 Portable hardness tester for variety of steel testing.



Photo 143 Shore Schleroscope hardness tester, particularly accurate for larger pieces.

The predecessor of the modern Loeb system, still of practical application in the home workshop, is the original Shore Schleroscope, which may sometimes be picked up as discarded industrial equipment. This was produced from the very start of the twentieth century, and can still provide a valuable means of hardness testing, although it requires relatively large test pieces of at least $\sim 1\text{in}^3/25\text{mm}^3$. As a simple mechanical test, it employs a hardened steel capsule dropped down a glass tube, from an air-bulb operated catch, the bounce height being observed directly in the tube. Rather tricky sighting of the rebound height can be enhanced by simple video photography, and a precise value for H_s obtained. The wide range of hardness that can be measured, combined with negligible permanent indent, are of benefit in the home workshop.

Finally, the good-old method of roughly assessing the effect of hardening a steel part by checking with a file, should be recalled. Whilst only a qualitative measure, the file check can provide a first indication of the success, or otherwise, of a steel hardening treatment. [Photo 144](#) shows commercially available files, to cover the range 45–65 H_{RC} , to provide a more quantitative indication of hardness, for minimal outlay.



Photo 144 Comparative hardness testing, for hard steels, with graded files from 40 to 65 HRC (Rockwell hardness, C scale).

Table 42 Hardness scale comparison and approximate steel (plain carbon and low alloy) strength.

Vickers (H_V)	Brinell (H_B)	Rockwell (H_{RC})	Shore (H_S)	UTS (ksi)	UTS (N/mm²)
760	600	62	80	293	2020
650	55	58	75	271	1870
572	514	55	70	250	1730
517	477	50	66	233	1600
475	444	48	61	217	1500
440	415	45	57	204	1405
411	388	42	53	188	1300
384	363	40	50	177	1220
361	341	38	47	166	1150
338	321	36	45	157	1080
318	302	34	42	148	1020
300	285	31	40	139	960
284	269	29	38	132	910
265	255	26	36	123	850
251	241	24	35	116	805
234	229	21	33	112	772
222	217		32	105	725
202	197		29	96	665
184	179		27	88	610
168	163		25	84	580
154	149		23	76	525
136	131		21	67	485
116	111		19	57	395
100	95			49	340

Appendix II: Resources

MODELLING MATERIALS, ENGINEERING SUPPLIES

Blackgates Engineering

Dewsbury, W. Yorks

Tel. 01924 466000

www.blackgates.co.uk

Macc Models

Macclesfield, Ches.

Tel. 0161 408 2938

www.maccmodels.co.uk

Stuart Models

Bridport, Dorset

Tel. 01308 456859

www.stuartmodels.com

Reeves 2000

Austley, Warks

Tel. 01827 830894

www.ajreeves.com

Hemingway Kits

Bridgenorth, Shrops

Tel. 01746 767739

www.hemingwaykits.com

Engineering supplies

Brabbin and Rudd

Bolton, Lancs

Tel. 01204 521171

www.brabbins-online.co.uk

Brian S. Pope

Stockport, Ches

Tel. 0161 480 8322

www.brianpope.com

Specialist materials

Ball bearings

Simply Bearings Ltd.

Leigh, Lancs

Tel. 01942 269837

www.simplybearings.co.uk

Rivets

Sapphire Products Ltd.

Nechells, W. Mids

Tel. 0121 326 6000

www.sapphireproducts.co.uk

Horseshoe nails

Arthur Cottam & Co. Ltd.

Chesterfield, Derbys

Tel. 01246 453672

www.cottamhorseshoes.com

Raw materials

Steels

Acenta Steels Ltd.

Rugby, Warks

Tel. 01788 535 352

www.acentasteel.com

KV Steels Services, Ltd.

Dalston, W.Mids

Tel. 0121 526 5666

www.kvsteel.co.uk

Ferrous & Non-Ferrous

College Engineering Supplies

Wolverhampton, W. Mids

Tel. 01902 843384

www.collegeengineering.co.uk

Holme Dodsworth

Newcastle-upon-Tyne

Tel. 0191 265 9077

www.holmedodsworth.com

Smith Metal Centres

Leeds, W.Yorks

Tel. 01767 604704

Casting equipment

John Winter & Co. Ltd.

Halifax, W. Yorks

Tel. 01422 364213

www.johnwinter.co.uk

Heat treatment

Temperature measurement

Maplin

(HO) Milton Keynes

Tel. 0843 227 7315 www.maplin.co.uk

Heat Treatment Salts (AS 220)

Techniques Surfaces

Erdington, W. Mids

Tel. 0121 382 6839

www.durferrit.com.

Small Furnaces/Kilns

Flamefast Ltd.

Manchester

Tel. 0161 793 9998

www.flamefast.co.uk

Corby Kilns Ltd.

Kettering, Northants

Tel. 01536 726361/521171

www.corbykilns.co.uk

Heat Treatment Services

Bodycote Ltd.

www.bodycote.com

Wallwork Heat Treatment Ltd.

Bury, Lancs

Tel. 0161 797 9111

www.wallworkht.com

Tooling and workshop equipment

Chronos Tools

Dunstable, Beds

Tel. 01582 471900

www.chronos.ltd.uk

Chester Tools

Chester, Cheshire

Tel. 01244 531631

www.chestermachinetools.com

RDG Tools

Mytholmroyd, W.Yorks.

Tel. 01422 885069

www.rdgtools.co.uk

Machine Mart

(HO) Nottingham

Tel. 0844 880 1250

www.machinemart.co.uk

Appendix III

ELEMENTS

Periodic table no.	Symbol	Element
4	Be	beryllium
5	B	boron
6	C	carbon
11	Na	sodium
12	Mg	magnesium
13	Al	aluminium
14	Si	silicon
15	P	phosphorous
16	S	sulphur
17	Cl	chlorine
18	Ar	argon
19	K	potassium
20	Ca	calcium
22	Ti	titanium
23	V	vanadium

24	Cr	chromium
25	Mn	manganese
26	Fe	iron
27	Co	cobalt
28	Ni	nickel
29	Cu	copper
30	Zn	zinc
40	Zr	zirconium
41	Nb	niobium
42	Mo	molybdenum
46	Pd	palladium
47	Ag	silver
48	Cd	cadmium
50	Sn	tin
51	Sb	antimony
73	Ta	tantalum
74	W	tungsten
78	Pt	platinum
79	Au	gold
80	Hg	mercury
82	Pb	lead

ABBREVIATIONS

Abbreviation	
AISI	American Iron and Steel Institute
ASTM	American Society for Testing and Materials
BS	British Standards
EN	prefix for 1955 BS steels standards
$A_s; A_f$	Austenite start temperature; Austenite finish temperature (TTT diagrams)
$B_s; B_f$	Bainite start and finish, as austenite
$M_s; M_f$	Martensite start and finish, as austenite
$A_{C1}; A_{R1}$	steel equilibrium diagram, lower critical temperature (1) on heating (C); on cooling (R)
$A_{C3}; A_{R3}$	steel equilibrium diagram, upper critical temperature (3) on heating (C); on cooling (R)
BM	bending moment (e.g. in beam bending)
SF	shear force (e.g. in beam bending)
UDL	uniformly distributed load (e.g. in beam bending)
CE	Carbon Equivalent (measure of weldability)
CP	commercial purity
TTT	time, temperature, transformation diagram
CCT	continuous cooling, transformation diagram
PS	proof stress, usually yield at 0.2% strain

YP	yield point, point of deviation from elastic to plastic strain
UTS	ultimate tensile stress
Qu & T	quench and temper (e.g. steels)
T _m	melting point
H _V H _B H _R H _S	Hardness – Vickers, Brinell, Rockwell, Shore
TEM	transmission electron microscope
SEM	scanning electron microscope
STEM	scanning tunnelling electron microscope
psi	stress and pressure lbf / in ² Imperial units
ksi	lbf / in ² × 10 ³
N/mm ²	stress and pressure SI units
MN/m ²	1 N/mm ² = 1 MN/m ² = 1 MNm ⁻²
MNm ^{-3/2}	usual units of fracture toughness
mm	millimetre (mm) = metre (m) × 10 ⁻³
µm	micrometre = metre (m) × 10 ⁻⁶
nm	Nanometre = m × 10 ⁻⁹
N	Newton = kg / 9.81
kN	(kilo) N × 10 ³
MN	(mega) N × 10 ⁶
GN	(giga) N × 10 ⁹
da/dN	fatigue crack growth rate (e.g. mm / cycle)

N	number of fatigue cycles
HAZ	heat affected zone (adjacent to weld)
PM	parent metal
WM	weld metal
MMA	(SMAW – USA) /)‘stick’ / Manual Metal Arc welding
MIG	(GMAW – USA) /)‘semi-auto’ / Metal Inert Gas welding
TIG	(GTAW – USA) Tungsten Inert Gas welding

SYMBOLS

Symbol	
σ	stress (units eg. N/mm ²); σ_y yield stress ; σ_u ultimate tensile stress; σ_R radial stress; σ_L longitudinal stress; σ_m mean stress
ϵ	strain (unit-less); micro-strain ($\mu\epsilon$) = $\epsilon \times 10^{-6}$
ν	Poisson’s ratio (unit-less)
τ	shear stress
δ	deflection (beam bending); or phase in equilibrium diagram
α	alpha phase (often ferrite) in equilibrium diagram
β	beta phase in equilibrium diagram
γ	gamma phase (often austenite in equilibrium diagram)

Y	Specimen width correction factor (fracture mechanics)
Σ	Summation (e.g. $\Sigma 1,2,3,4,5 = 15$)
Δ	change in, e.g. ΔL = change in length)
E	Young's Modulus (e.g. N/mm ²)
G	Modulus of Rigidity (e.g. N/mm ²)
I	Second Moment of Area (e.g. mm ⁴)
J	Polar Second Moment of Area (e.g. mm ⁴)
$^{\circ}\text{K}$	degree Kelvin ($^{\circ}\text{K}$); $0^{\circ}\text{K} = -273^{\circ}\text{C}$
K_I	fracture toughness, mode I opening (MNm ^{-3/2})
K_{III}	fracture toughness, mode III opening (MNm ^{-3/2})
K_{IC}	plain strain fracture toughness
K_t	stress concentration factor

CONVERSIONS

Unit	Equivalent
length	
1 in (inch)	25.4mm
mass	
1 kg	2.205 lb
1 tonne	1000 kg

1 ton	2240 lb
force	
1 kgf	9.81 N; 2.205 lbf
stress	
1 N/mm ²	1 Nmm ⁻² ; 1MNm ⁻² ; 145lbf/in ² ; 145 psi; 0.145 ksi
temperature	
0°C	273°K
density	
1 g/mm ³	0.03613 lb/in ³

Note: this book unashamedly contains a mixture of basically Imperial and SI (metric) units, which whilst being somewhat messy, does reflect practice and there is a case to be made for being conversant with both systems, rather like my favourite Boxford lathe with its mm graduated saddle tumblers and Imperial tailstock and screwcutting features, which works perfectly well despite such an apparently shambolic arrangement. This is also reflected in international practice, with universal metrication in continental Europe, Asia and Australasia, near universal Imperial use in the USA, and a complete mixture in the UK!

Glossary

annealing	heat treatment to reduce hardness and increase ductility
ausforming	elevated temperature working based on phase and TTT diagrams
austempering	heat treatment by cooling at controlled rate for preferred phases
austenite	metallic phase, often fcc
bainite	phase detailed on TTT diagram
case hardening	diffusion of carbon in steel surface for subsequent hardening
castability	measure of molten characteristic to form sound casting
cementite	Fe_3C , steel phase
Charpy, Izod	impact (toughness) testing methods
columnar (grains)	long grains formed on solidification, near surface
crystal structure	basic atomic structure, e.g. face centred cubic
dendritic	solidification structure in casting, fir-tree like
dislocation	discontinuity, fault in crystal assembly from perfect array
equiaxed (grains)	randomly arranged, small grains preferred
fatigue	effect on metal structure of cyclic loading
fatigue life	number of cycles of fatigue loading to failure

fatigue limit	stress which metal can sustain without fatigue failure
galvanizing	zinc coating of steel base metal, for galvanic protection
heat treatment	process of heat application to modify structure as required
ingot	casting product, usually for further processing
martensite	phase formed in diffusion-less manner during heat treatment
martempering	working of a steel above the M_s , before controlled cooling
Nitriding	proprietary hardening process of steel, diffusing nitrogen into surface
normalizing	heat treatment to obtain preferred balance of toughness and strength
pearlite	lamellar structure of ferrite and cementite, in steels
precipitation hardening	Re-heating of quenched alloy for increased hardness, ageing process
quenching	rapid cooling to achieve desired phase(s)
recrystallization	process of micro-structural improvement from cold work and heat treatment
solution treatment	heat treatment, usually referring to non-ferrous alloys
strain	measure of change in length
stress	effect of loading on micro or macro-structure
supersaturated solution	arrested transformation with excess phase with insufficient time to achieve equilibrium structure
tempering	Re-heating process to relieve as-quenched hardness

threshold stress	stress below which material is safe from fatigue failure or crack growth
weldability	Measure of ability to achieve full strength fusion weld
yield point	point of deviation from linear elastic stress-strain behaviour

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