

STRENGTH MIS-MATCH EFFECT ON FRACTURE BEHAVIOR OF STRUCTURAL STEEL WELDS

77241

**A Thesis Submitted to the
Graduate School of Natural and Applied Sciences of
Dokuz Eylül University
In Partial Fulfillment of the Requirements for
the Degree of Doctor of Philosophy in Mechanical Engineering, Mechanics
Program**

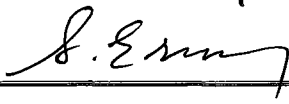
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**April, 1998
İZMİR**

Ph. D. THESIS EXAMINATION RESULT FORM

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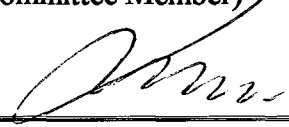
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ACKNOWLEDGEMENTS

First of all, I would like to thank Assoc. Prof. Dr. S. Erim, my advisor, whose guidance, constructive suggestions and continued interest were vital to the completion of this work. It is really a chance to study under his supervision.

This work was part of a joint project between Dokuz Eylül University and GKSS Research Center, Germany. It has been financially supported by TÜBİTAK and International Büro Jülich, Germany. I wish to express my deep indebtedness to these Institutions for their financial support and endless contributions.

The experimental and literature review parts of this work has been carried out at GKSS, Research Center, Institute of Materials Research, Geesthacht, Germany. I would like to gratefully acknowledge the helpful encouragement and guidance received throughout the course of this study from Dr. M. Koçak, Head of the Welding Group and my supervisor. I had the benefit of reviews, comments and advises from Dr. G. Çam. I express my special thanks for his constructive suggestions, guidance and his endless support in my work. I would like to thank Prof. Dr. K.-H. Schwalbe, Director of the Institute of Materials Research, for letting me conduct this research. Members of the staff at the Institute were very helpful and I would especially like to thank Mr. S. Riekehr, Mr. J. Knaack and Mr. H. Mackel for their help in conducting the tests.

I also wish to express my gratitude to my colleagues at the Department of Mechanical Engineering, especially Res. Ass. K. Şener, for their help and support.

Finally, I would like to thank my family for their love, encouragement and constant moral support; their contribution makes it all worthwhile.

Çınar YENİ, İzmir, 1998

ABSTRACT

For the welding of many structural metallic materials, laser beam (LB) offers many advantages over traditional welding processes. The advantages include a low heat input, narrow bead width and heat affected zone (HAZ), low distortion, welding in an exact and reproducible manner at high speed. However, the mechanical testing methods of laser welded similar and dissimilar joints and the response of conventional materials to this new process have not yet been fully established. Laser welded joints are not yet commercially used in Turkey. The use of such joints are in the experimental phase. The aims of this project are to gain more information on the heterogeneity of the welded region by means of experimental procedures, following the fracture mechanics methods, to help establish the mechanical and metallurgical properties of laser welded joints from the fracture mechanics aspects and also help to set-up a standard test procedure method for laser welded joints. It is strongly believed by many researchers that laser welding which offers possibilities for automotization and robotization for welding to be carried out in areas of difficult access in severe operating conditions, among its many other advantages will be one of the most widely accepted welding techniques worldwide in near future.

This study was undertaken to investigate the microstructural evaluation occurring during welding and to determine the mechanical properties and fracture behavior of laser welded similar and dissimilar ferritic and austenitic steel joints. A complete information on the tensile and fracture toughness properties of the fusion zone is essential for pre-qualification and understanding of the joint performance in service as well as for conducting the defect assessment procedure for such very narrow weld joints. An experimental investigation on the mechanical properties of laser weld joints were carried out to establish a testing procedure using flat microtensile specimens for determination of the tensile properties of the weld metal and heat

affected zone (HAZ) of the laser beam welds. These results are compared to those obtained with standard transverse tensile and round tensile tests. To accomplish this, three similar and two dissimilar butt-joints were produced by CO₂ laser, using 6 mm thick steel plates. The mechanical properties of laser welded specimens have been further examined by extensive microhardness measurements and microstructural evaluation has been carried out on the joints. The CTOD (Crack Tip Opening Displacement) and CMOD (Crack Mouth Opening Displacement) tests were performed at Room Temperature and -40 °C, to determine the fracture toughness of similar and dissimilar LB welds and the variation of the fracture behavior with temperature by using three point bend specimens. The effect of mechanical heterogeneity (mis-matching) at the vicinity of the crack tip on the CTOD fracture toughness values was also discussed.

This study is part of a joint project between Dokuz Eylül University and GKSS Research Center, Germany on the research on LB welds carried out at GKSS Research Center. It is financially supported by TÜBİTAK, Turkey and International Bureau Jülich, Germany.

ÖZET

Birçok metalik malzemenin kaynağında lazer ışını (LB) klasik kaynak yöntemleriyle karşılaştırıldığında oldukça fazla avantajlara sahiptir. Bu avantajlar arasında düşük ısı girdisi, dar kaynak dikişi ve ısıdan etkilenmiş bölge (HAZ), şekil bozukluğunun azlığı ve yüksek hızda düzgün kaynak sayılabilir. Buna rağmen, benzer ve farklı metalik bağlantıların mekanik test yöntemleri ve alışılmış malzemelerin bu yeni kaynak işlemine olan tepkileri henüz tam olarak belirlenmemiştir. Lazer kaynaklı bağlantılar Türkiye’de henüz ticari amaçlı olarak kullanılmamaktadır. Bu tür bağlantıların kullanımı deney amaçlı yapılabilmektedir. Bu projenin amaçları kırılma mekaniği yöntemlerini izleyerek deneysel yollarla kaynak bölgesinin heterojenitesi hakkında bilgi edinmek, kırılma mekaniği açısından lazer kaynaklı bağlantıların mekanik ve metalurjik özelliklerinin belirlenmesine yardımcı olmak ve ayrıca lazer kaynaklı bağlantılar için standard bir test yöntemi geliştirilmesine katkıda bulunmaktır. Birçok araştırmacı, erişilmesi güç yerlerde ve ağır dış ortam koşullarında kullanılabilen, otomatik ve robotize edilmiş bir kaynak yöntemi olan lazer kaynağının yakın gelecekte en çok tercih edilen ve kabul edilmiş tekniklerden biri olacağına inanmaktadır.

Bu çalışmada lazer kaynağı sırasında meydana gelen mikroyapı ve benzer ve farklı, ostenitik ve ferritik lazer kaynaklı bağlantıların mekanik özellikleri ve kırılma davranışları incelenmiştir. Kaynak bölgesinin çekme ve kırılma tokluğu özelliklerinin tam olarak belirlenmesi, kaynağın yeterliliğinin ve bağlantı performansının servis çalışma durumunda tam olarak anlaşılabilmesi ve böylesine bir dar kaynak dikişine sahip bağlantılar için hata belirleme prosedürlerinin uygulanabilmesi açısından son derece önemlidir. Lazer kaynağı yapılmış bağlantıların kaynak ve ısıdan etkilenmiş bölgelerinin mekanik özelliklerinin belirlenebilmesi amacıyla düz mikroçekme

numuneleri kullanılarak çekme deneyleri yapılmıştır. Elde edilen değerler standart çekme ve yuvarlak çekme numuneleri ile elde edilen değerlerle karşılaştırılmıştır. Bunun için 6 mm kalınlığında çelik plakalardan üç benzer ve iki farklı CO₂ lazer kaynaklı bağlantı elde edilmiştir. Mekanik özellikler ayrıca mikroyapı incelemeleri ve mikrosertlik ölçümleri ile de araştırılmıştır. Benzer ve farklı lazer kaynaklı bağlantıların kırılma tokluklarının ve kırılma davranışlarının sıcaklıkla değişiminin belirlenebilmesi için oda sıcaklığında ve -40 °C’de üç noktalı eğilme numuneleri kullanılarak Çatlak Ucu Açılma (CTOD) ve Çatlak Ağızı Açılma (CMOD) ölçümleri yapılmıştır. Çatlak ucu bölgesindeki mekanik heterojenitenin (mis-match) CTOD üzerindeki etkisi araştırılmıştır.

Bu çalışma Dokuz Eylül Üniversitesi ile Almanya, GKSS Araştırma Merkezi arasında Lazer Kaynağı konusunda yürütülen ortak projenin bir bölümünü oluşturmaktadır. Türkiye’den TÜBİTAK, Almanya’dan Jülich Araştırma Merkezi tarafından parasal olarak desteklenmektedir.

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CHAPTER ONE

INTRODUCTION

1.1 Introduction

Although laser beam (LB) welding is a new process for joining materials, it has potential applications in various industries, such as aerospace, shipbuilding and automotive, both in macro- and microstructural applications. This high power density welding process has unique advantages of cost effectiveness, low distortion, high welding speed, easy automation, deep penetration, narrow bead width and narrow HAZ compared to the conventional fusion welding processes. Usually there is no need for filler metals, there is low distortion and consequently very little post weld straightening (Mazumder, 1993). Despite the high capital cost of laser weld equipment, it is expected that laser welding will make a great impact on welding of structural and high precision parts in various manufacturing and maintenance industries. It is currently used in the welding of a variety of metals and alloys including hot and cold rolled steels, high strength low alloy and stainless steels, aluminum alloys, titanium alloys, refractory and high temperature alloys and dissimilar materials, primarily in thin section areas. It is, however, foreseen that within the next decade laser welding will break through into the industrial areas for the welding of medium section steel structures. Especially in the area of constructional steels the application of laser welding will have a tremendous impact on the whole concept of fabrication. The very high accuracy to which the different elements can be welded, will have a huge influence on the methods of assembly. The process will further make it possible to introduce new design concepts, which up to now have been impossible to follow due to the limitations connected with traditional welding techniques. The reason for this is threefold. Firstly, there is an obvious need

for a new welding process due to the inherent limitations to the existing welding processes with regard to quality speed, productivity and fitness for automation. Secondly, robust high power lasers suitable for such thicknesses have now been developed, and, thirdly, laser welding offers many advantages compared to arc welding as mentioned before (Metzbower, 1988). However, structural significance of the metallurgical and mechanical properties of LB welds combined with the unique shape of the weld joint is still need to be established for structural components. Determination of the metallurgical properties and fracture mechanics behavior of laser welded similar and dissimilar materials is the subject of many research projects carried out worldwide since laser welded joints are beginning to find wide acceptance in industrial applications.

Laser welded joints like all other weld joints may contain defects in the form of cracks in the narrow weld area or may develop from inhomogeneities during service. The size and location of such cracks associated with characteristics of the local microstructure and the level of mechanical strength mis-match between weld zone and parent plate under certain service condition directly affects joint performance/life time of a structure (Sun & Moisio, 1994). It is now known that the variation in mechanical properties of different regions (strength mis-match) of a conventional weld joint significantly influence the deformation characteristics of welds in service (Koçak & Junghans 1994), (Cao & Evans, 1989). Laser beam welding produces a very narrow weld bead and HAZ, which may be an advantage over conventional processes if weld joint shows high hardness/strength (i.e. low toughness) due to the occurrence of plastic deformation in the parent plate under external stresses (Johnson, 1988). However, for pre-qualification or defect assessment purposes, the determination of the mechanical properties of very narrow LB fusion zone cannot be done by standard testing procedures. For instance, it is usually observed that transverse tensile specimens fail in the softer base metal and fracture path deviation towards the base metal occurs in Charpy-V and fracture mechanics based crack tip opening displacement (CTOD) testing of C-Mn steel welds due to mechanical property mis-match between the base metal and the weld zone (Yeni, Erim, Çam &

Koçak, 1996), (Çam, Riekehr & Koçak, 1997). Therefore, all these standardized test results may provide information on the performance of whole joint under tensile, impact and bending loading conditions, respectively. They may fail in providing required intrinsic tensile and toughness properties of the fusion zone due to interaction (unavoidable) between weld zone and parent plate under applied deformation. For the CTOD toughness determination, the effects of specimen geometry (e.g. weld width, crack size, notch position, etc.) and the degree of strength mis-match between base metal and weld zone on toughness have to be taken into account (Koçak & Schwalbe, 1994), (Kristensen & Borggreen, 1994), (Koçak, Es-Souni, Chen & Schwalbe, 1989), (Kirk & Doods, 1992).

The high speed laser welding process may lead to drastic changes in the microstructure with accompanying effects on the mechanical properties and hence on the performance of the joint. The thermal cycles associated with laser beam welding are generally much faster than those involved in the conventional arc welding processes. It does, however, mean that the rather small weld zone will locally exhibit a high hardness for C-Mn structural steels due to the formation of martensite and other fast thermal cycle constituents. The mechanical properties of these welds such as hardness, ductility and toughness are, therefore, a matter of prime concern. Furthermore, precise and easy evaluation of these properties for quality control is critical for a wide scale application of this process (Metzbower, 1990).

In order to determine the metallurgical and mechanical properties/fracture behavior of laser welded materials, three different base metals, namely two ferritic St37 and St52 and one austenitic 1.4404 were chosen. They were laser welded in five different combinations. Three of them are similar (St52-St52), (St37-St37), (1.4404-1.4404) and the other two are dissimilar (St52-1.4404), (St37-1.4404). The joints were produced by CO₂ laser without using filler wire.

Round tensile specimens and standard flat transverse tensile specimens were extracted from the welded plates and tested at room temperature. In order to obtain

the local mechanical properties of the welded joints at the base metal (BM), weld metal (WM) and heat affected zone (HAZ) regions, sets of flat microtensile specimens (0.5 mm thick, 2 mm wide) were extracted and tensile tests were carried out at room temperature. The obtained results were presented in diagrams and the two tensile test results were compared with each other. The suitability of the flat microtensile specimen testing procedure for laser welds in determining the tensile property gradient existing in the fusion zones of the laser beam welded similar and dissimilar steel joints is confirmed. The fracture surfaces of some selected specimens were examined by scanning electron microscopy. In addition to the microhardness measurements performed across the welded regions, microstructural examinations of the joints were also carried out. Related diagrams and micrographs have been presented in the following chapters.

Single edge notch bend (SENB) specimens, machine notched in the weld zone were fatigue pre-cracked in order to introduce a sharp crack. Crack Tip Opening Displacement (CTOD) measurements using the GKSS developed δ_5 clip gauges (GKSS-Displacement Gauge Systems for Application in Fracture Mechanics) and Crack Mouth Opening Displacement (CMOD) measurements were conducted at two different temperatures, namely at room temperature and $-40\text{ }^{\circ}\text{C}$. Comparison of the results with regards to joint type and test temperature have been given in the related Tables and diagrams. At the final stage SENB specimens were broken by further fatigue cracking. In order to examine the fracture surfaces, broken specimens were cut near the fracture surfaces.

Selected specimens from each weld group were photoelastically coated in order to determine the deformation patterns when the specimen is subjected to tensile or bending loading. Photographs demonstrating the deformation patterns are presented in the related chapter.

The obtained results are discussed in terms of microstructure, hardness, mechanical properties and fracture behavior. These results were used to clarify the

effect of mechanical heterogeneity (mis-match) properties and fracture behavior. It is expected that this project will form one of the first steps in the determination of metallurgical and mechanical properties of laser welded similar and dissimilar materials, taking into account the extreme mechanical heterogeneity (in terms of strength values) between the base metal and the weld region. It is believed that more information will be obtained on this heterogeneity by means of experimental and theoretical procedures. Further tests on laser welded similar and dissimilar various materials will help to gain an understanding of the mechanical property/fracture mechanics behavior and the microstructures of laser welded materials.

Specific objectives of this study include:

- 1) Preparing a literature review on laser beam welding and strength mis-match, especially with emphasis on ferritic and austenitic steels.
- 2) Investigating microstructural evolution in CO₂ laser welded similar and dissimilar ferritic and austenitic structural steels.
- 3) Illustrating shortcomings in determination of mechanical properties of joints by standard transverse tensile testing.
- 4) Determining the mechanical properties (yield strength, tensile strength, percent elongation, etc.) across the joints by using flat microtensile specimen technique.
- 5) Correlating microstructural/mechanical properties of weld regions of joints.
- 6) Determining the hardness profiles of the laser welded specimens.
- 7) Evaluating the effect of mechanical heterogeneity (mis-match) on deformation behavior of joints.
- 8) Obtaining the crack tip opening displacement (CTOD) and crack mouth opening displacement (CMOD) curves of the base metal and the weld region at room temperature and -40 °C by multispecimen method.
- 9) Evaluating the effect of temperature on the fracture toughness values.
- 10) Help establish a standard test procedure for the testing of laser welded similar and dissimilar materials.

CHAPTER TWO

LITERATURE REVIEW

2.1 Laser Beam Welding

2.1.1 Definition and Development

The joining of any material with another has always been a necessity in engineering applications. Welding or bonding of two materials always caused changes or deterioration of the microstructural and mechanical properties of the original materials at the joining region. Joining of two different materials (dissimilar joints) leads a much more complicated situation where each material's own properties should be taken into account during the joining process and also later on its joint quality determination and also later on, on its joint performance in service.

In addition to the conventional heat sources used in fusion welding processes, namely gas flame and arc, electromagnetically accelerated particle beams such as electron, ion and neutral beams and special type beams (i.e. laser and plasma beams) with large outputs developed over the last 30 years are available which are named as “power beam welding” procedures (Dilthey, Fuest, Huwer, & Jacobskötter, 1991).

The **power beam welding** processes are generally classified as those which are capable of delivering an energy density of greater than $\approx 10 \text{ kW/mm}^2$ to the workpiece, and which are thereby capable of deep ‘keyhole’ penetration. These processes comprise electron beam, plasma arc and laser welding (Sandia National Laboratories, 1996).

Electron beams which are easy to generate and offer relatively higher output were the first high energy density beam to be commercially applied. Its most

important disadvantage is the requirement of a vacuum working chamber since its propagation range in air is extremely short. However, it is possible to weld much thicker plates by electron beam welding process (Schulze & Schubert, 1997).

The plasma arc technique is the only arc welding process which is capable of achieving the energy density required for 'keyhole' penetration. In this technique, an arc is formed between a tungsten electrode and the workpiece, but is constricted and intensified by a copper nozzle surrounding the electrode and, giving a plasma arc of a columnar form. The combination of a high welding current and plasma gas flow rates results in deep penetration of the workpiece. The plasma arc technique is not only capable of high penetration keyhole welding; it can also be used as a low penetration weld surfacing technique procedures (Dilthey et al., 1991).

Laser Beam Welding (LBW) uses a moving high density (10^5 to 10^7 W/cm²) coherent optical energy source called a laser as the source of heat. "Laser" is an acronym for "light amplification by simulated emission of radiation". The coherent nature of laser beam allows it to be focused to a small spot, leading to high energy densities. A laser beam is generated when photons traveling in a direction along the longitudinal axis of the crystal are reflected and returned to the crystal by the end mirrors where they continue to amplify the output through the phenomena of stimulated emission. Since the front mirror is up to 60% light transitive a laser beam is emitted. This beam of light is monochromatic, collimated and coherent. Because of these three characteristics and resultant low divergence, the beam is capable of traveling great distances with minimal energy loss.

Lasers have been promoted as potentially useful welding tools for a variety of applications. Until the 1970's, however, laser welding had been restricted to relatively thin materials and low speeds because of the limited continuous power available. By 1965, a variety of laser systems had been developed for making microwelds in electronic circuit boards, inside vacuum tubes and in other specialized

applications where conventional technology was unable to provide reliable joining (Mazumder, 1993).

Lasers may be classified according to their mode of operations either as continuous wave (CW) or pulsed wave (PW) lasers. Early lasers capable of melting and cutting materials were of the solid state type with a pulsed output whereas the continuous lasers did not have sufficient power output for processing the materials. However, the development of the high power continuous wave (CW) CO₂ gas lasers (wave length of 10.6 μm) and the pulsed Nd:YAG (neodymium doped Yttrium Aluminum Garnet) lasers (wave length of 1.06 μm) in the last two decades led to the application of the laser beam as the more general heat source in areas, such as cutting, welding, heat treating, soldering, brazing, cladding and hardfacing. Fig. 2.1 illustrates a laser beam welding operation (Energy Beam Labs Reference Center, 1997) and Fig. 2.2 presents the principles of laser beam welding (Scharfe, 1993).

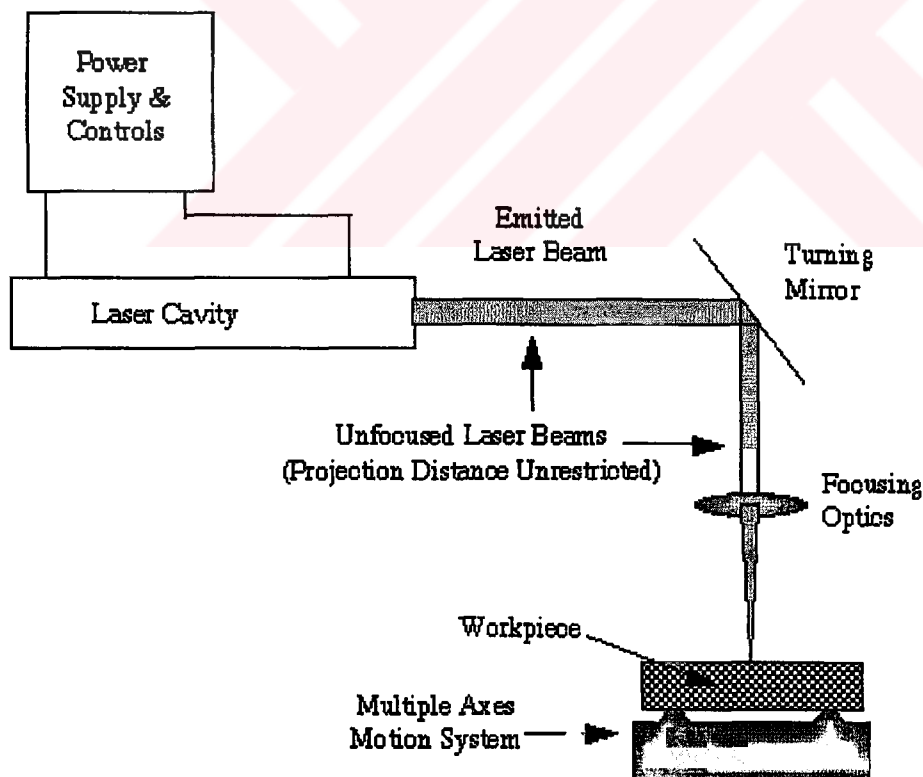


Fig. 2.1 Schematic illustration of a laser beam operation

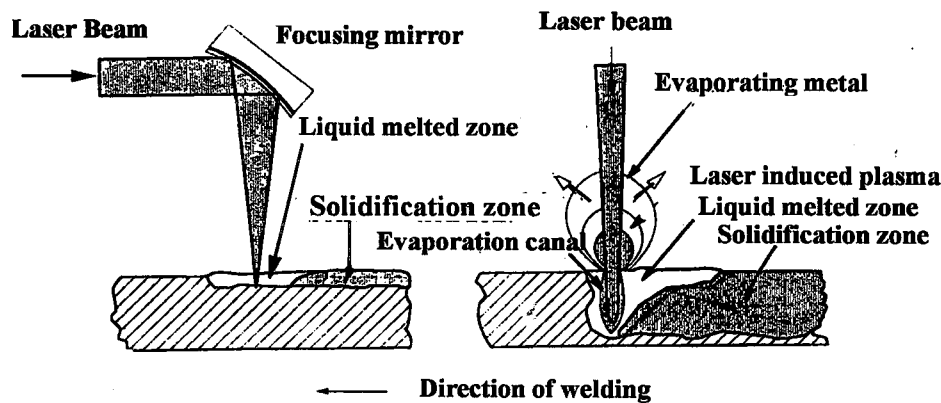


Fig. 2.2 Principles of laser beam welding

2.1.2 Properties

In the industrial processing of materials, lasers are used to cut, weld, drill, etch, perforate and heat-treat a variety of substances. Laser is an ideal source for these tasks because it delivers appropriate thermal energy in a tightly focused beam. Flexible and easy manipulated, laser beam can be optimally shaped to direct intense laser power, accurately and repeatedly, on almost any target or work surface (Roland, Lembeck, Jaudas & Lankenau, 1996).

Laser beam welding is a manufacturing process which has advantages such as deep weld penetration, high speed, minimized heat input, rapid cooling and narrow weld region. It requires little or no filler materials with very little distortion of the joint. Metals that have inherent magnetic properties can be welded. It is highly suited for automation by precisely positioning with optical elements that are located remote from the fusion zone. It can be transmitted substantial distances through the atmosphere without serious attenuation or optical degradation because of its coherent nature, which provides the opportunity for multi-station working. Laser focusing and beam delivery optics are readily incorporated in computer programmed, multi-axes motion systems. High volume production assemblies can be manipulated under the programmed welding optics. Also, the laser emits a beam of light as its welding energy, free of restraining cables, electrical connections and arc length limitations (Leong & Hunter, 1996).

Since laser welding is done at atmosphere, loading and unloading of production quantities are often done fast and therefore, the requirement for complex tooling to fit into a vacuum chamber (as in the electron beam process) is avoided. Laser beam welding is flexible because the beam can be moved under digital control to seam weld any shape. Difficult-to-reach areas can be accessed by reflection from mirrors and fiber optics. Laser beam welding thus offers easily maneuverable, chemically clean, high intensity, atmospheric welding process with narrow HAZ and subsequent low distortion (Sekhar, Robinson & Reed, 1996).

Laser welding specifically offers the following advantages:

- Light is inertialess (hence, high processing speeds with very rapid stopping and starting become possible).
- Focused laser light provides high energy density.
- Laser welding can be used at room atmosphere.
- Difficult-to-weld materials (for example titanium, quartz, etc.) can be joined.
- Workpieces do not need to be rigidly held.
- No electrode or filler materials are required.
- Narrow welds can be made.
- Precise welds (relative to position, diameter and penetration) can be obtained.
- Welds with little or no contamination can be produced.
- The heat affected zone (HAZ) adjacent to the weld is very narrow.
- Intricate shapes can be cut or welded at high speed using automatically controlled light deflection techniques.
- Laser beam can also be time shared (Mazumder, 1993).

2.1.3 Comparison to Other Welding Procedures

Laser beam welding has been developed as a necessity. The reason for this is that there is an obvious need for a new welding process in major industrial areas due to the inherent limitations to the existing welding processes with regard to distortion, quality, productivity, fitness for automation, health and safety. The two dimensional heat distribution and consequently wide and shallow molten pool of arc welds means,

that many properties as speed and heat input generally speaking cannot be improved further due to the limitations caused by the physics of the processes. The distortion observed in combination with arc welding is for the same reasons always considerably, and this is a serious problem in the later stages of assembly. Of equal importance are the big difficulties met in the robotization of the processes. The control of the individual weld-pass geometry is a key problem to automation and robotization in arc welding. The number of independent variables is extremely large and the physical phenomena very complex. Because of these limits faced in traditional welding laser beam welding has emerged as a new joining process. Laser welding possesses most of the demanding properties like high speed, low distortion, single pass welding in large thicknesses and easy automation and robotization. It also has the advantage of being able to operate in any desired atmosphere (Mazumder, 1993).

Energy consumption and efficiency of laser beam welding compared to some other welding processes are given in the following Table (Energy Beam Labs Reference Center, 1997).

Table 2.1 Energy consumption and efficiency of LBW relative to other selected welding processes

Welding Process	Intensity of energy source (W/cm ²)	Joining efficiency (mm ² /kJ)	Fusion zone profile
Oxyacetylene (OAW)	10 ² -10 ³	0.2-0.5	Shallow for single pass
Arc welding	5x10 ² -10 ⁴	0.8-2 (a) 2-3 (b) 4-10 (c)	Shallow for single pass
Plasma arc (PAW)	10 ³ -10 ⁶	5-10	Shallow at low energy end Deep penetration at high energy end
Laser Beam	10 ⁵ -10 ³	15-25	Shallow at low energy density range Deep penetration at high energy density range
Electron beam	10 ⁵ -10 ⁸	20-30	Deep penetration
(a) Gas tungsten arc welding (GTAW) (b) Gas metal arc welding (GMAW) (c) Submerged arc welding (SAW)			

This high power density process brings about some limitations, however. The laser's most significant limitation is weld penetration. Because weld joint penetration is dependent on conducted heat, the thickness of the materials to be welded is not as thick as those obtained by conventional techniques. Laser keyholing is limited to about 20-40 mm thickness depending on the laser power and for these depths multi-kilowatt lasers must be utilized. Also the precise part fit-up and alignment are much more critical in laser welding than in ordinary arc welding. The typical focal spot diameter for a laser beam ranges from 100 to 1000 μm (Ion, Salminen & Sun, 1996).

In addition, procedures for laser welding have only been developed for a limited number of specific applications. The data required in procedure development must be established by experimental testing, a lengthy and expensive process hindered by the complex interactions which occur between the numerous process variables.

Moreover, the laser beam process does give rise to special considerations with regard to the mechanical/microstructural properties. The weld is very narrow, wherefore the actual evaluation of the mechanical/materials properties implies a number of challenges. The actual evaluation of the properties is problematic, as the established techniques cannot be used without modifications. Precise and easy evaluation of mechanical/material properties for quality control is mandatory.

Finally, the capital cost for laser welding devices is almost 10 times more expensive than comparable arc welding systems, though the operative costs are comparable or lower. Therefore, laser welding's significantly higher production speeds can justify the higher capital cost when compared with arc welding processes (Mazumder, 1993).

A comparison of laser beam welding to other welding processes with regards to property, cost and operational aspects is given in the following Table (Johnson, 1988).

Table 2.2 Comparison of Welding Processes

	Laser	TIG	Plasma	Electron Beam
Power absorbed by the workpiece	4 kW	2 kW	4 kW	5 kW
Total power used	50 kW	3 kW	6 kW	6 kW
Typical traverse speed	16 mm/s	2 mm/s	5.7 mm/s	40 mm/s
Alignment accuracy required	±0.5 mm	±1.0 mm	±1.0 mm	±0.3 mm
Energy input per unit length (absorbed by the workpiece)	250 J/mm	1000 J/mm	600 J/mm	125 J/mm
Capital cost and availability	\$250,000 for 5 kW output, available in one year delivery in UK and USA. \$750,000 for 15 kW output, available on one year delivery in USA.	\$16,000 welding set, immediate delivery	\$20,000 per welding set, immediate delivery	\$100,000 for 6 kW machine, on short delivery
Possibility of all positional welding	Yes but requires optics to manipulate the beam Optimum when moving workpiece	Serious penetration characteristic changes with attitude	Serious penetration characteristic changes with attitude	Yes but requires mechanics to move gun, optimum when moving workpiece
Distortion: Axial shrinkage	Minimal (small HAZ)	Significant ~1 mm on 5 mm plate	Significant ~1 mm on 5 mm plate	Minimal (small HAZ)
Angular	Minimal parallel sidedness	Significant V-shaped weld	Significant V-shaped weld	Minimal parallel sidedness
Surface profile defects	Very fine flow lines	Underside protrusion held in by surface tension	Underside protrusion held in by surface tension	Produces ruffled swarf on backface
Special requirements for process operation	Safety interlock to guard against misplaced beam reflections	Normal light screening	Normal light screening	Vacuum chamber, local vacuum, x-ray screening
Process end effects:				
Start	Slight surface protrusion	Smooth	Slight surface protrusion	Slight surface protrusion
Finish	Smooth	Smooth	Slight surface protrusion	Slight surface protrusion

It is believed that research projects carried out worldwide on laser beam welding will allow for a wider use of this process in industrial applications and in time a significant percent of the limitations will be overcome.

2.1.4 Joint Fit-Up

Because laser welds are very narrow the fit-up of the workpieces is important if autogenous welds are to be made. When welding sheet metal joints, both joint face gaps and surface fit-up have to be considered. Gaps at the joint face present the biggest problem since, if they are larger than the beam focused spot size the beam will pass straight through the joint. The thicker the material the larger the gap that can be welded, as there is more bulk material for bridging which results from the metal slumping from the sides of the joint (Hall & Wallach, 1995). Fig. 2. 3 shows some joint designs for laser beam welds on sheet metal (Shewell, 1977).

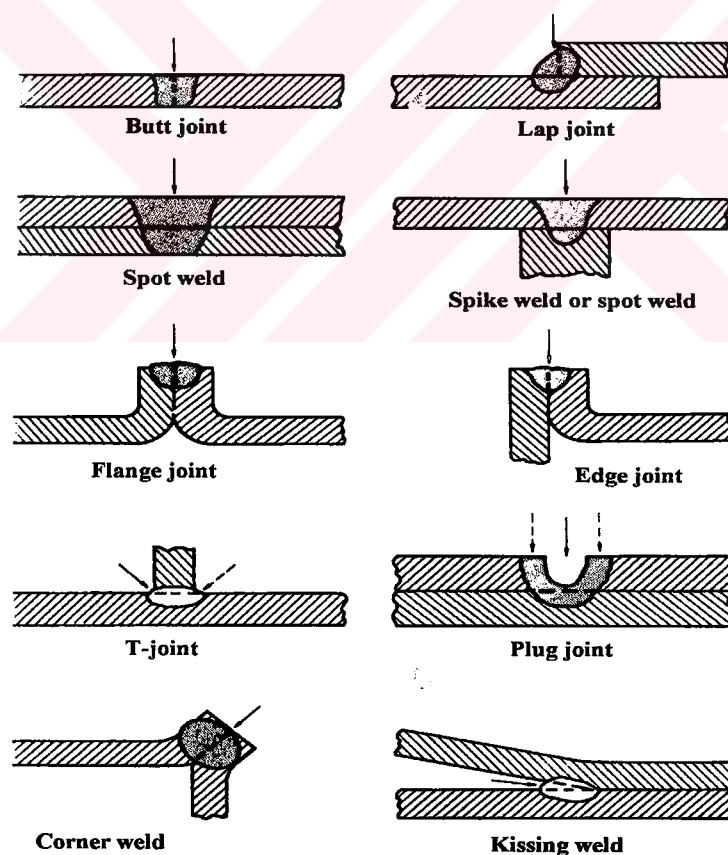


Fig. 2.3 Joint designs for laser beam welds on sheet metal. Arrows show direction of laser beam

2.1.5 Keyhole Formation

The laser beam generating temperatures up to 25000 °C rapidly brings metal to be joined to a boil around a hole filled with vapor, denoted as the “keyhole”.

The focal spot (thousandths of millimeters in diameters) is targeted on the weld joint surface or by focal length selection above or below it. At the surface the enormous concentration of light energy is converted to thermal energy. When the laser beam interacts with the workpiece, it delivers its heat to the surface and further penetration beneath the surface relies upon the thermal conduction. The metal under the laser beam evaporates and a cavity or keyhole is formed through the workpiece. By moving this keyhole along the joint between two pieces of metal, the molten metal solidifies, resulting in a fusion zone or weld. The hole acts as a blackbody, aiding the absorption of the laser beam as well as distributing the heat deep in the material. The keyhole mechanism and keyhole penetration can be seen in the schematic illustration in Fig. 2. 2.

2.1.6 Microstructure

Typical constructional steels produced on a modern steel work contain relatively low amounts of carbon compensated by the necessary amounts of other elements (primarily manganese) influencing strength and hardenability. The high cooling rates required to produce martensite in modern low carbon steels are not usually obtained in traditional arc welding and only seldom will martensite, therefore, be present in significant amounts. This situation is dramatically changed in high power, high speed laser welding of the same steels. The thermal cycle encountered here is sufficient fast such that often martensite and other fast thermal cycle constituents form. Depending on the actual thermal cycle, laser weld metal microstructures thus may contain appreciable amounts of martensite and bainite and may even be fully martensitic. Therefore, in laser beam welds, in contrast to arc welds, the weld metal dominates the properties of the joint. With regard to the steel composition, the obtained hardness values may be high, but at the same time there is a risk that ductility and toughness may be low (Kristensen, 1996).

2.1.7 Welding of Various Materials

All conventionally weldable materials can be laser welded. A range of dissimilar metals can also be joined, which were not previously weldable using arc fusion techniques. Examples of materials which are laser beam welded include:

- steel
- stainless steel
- aluminum
- titanium
- superalloys
- brass
- copper
- bi-metal
- spot welding
- differing thicknesses

Despite significant improvements in laser beam welding, there are still some complications in welding of these materials to be overcome. Here, particular emphasis will be given to steels since steel specimens were used in the experiments.

Presently, the major factor limiting the widespread use of laser beam welding process on steels is the certification/qualification of welded joints. The high cooling rate typical of this process promotes the formation of hard and brittle microstructures in a very narrow region within the weld region and HAZ of steels subjected to solid state phase transformations. Since the hardness values reached in this region are usually well above those specified in standards and codes, expensive and time consuming qualification procedures are required for individual components. Moreover, the evaluation of the mechanical properties in general is very difficult. First of all the actual measurement of the properties is difficult as established techniques cannot always be applied without some sort of modification (Yeni et al., 1996). Secondly, the acceptance criteria might have to be modified to accommodate

the weldment characteristics. For example the ductility cannot be measured by the standard tensile testing technique due to the narrow weld and the hard and high strength weld core. Impact toughness of the weld metal is often impossible to measure by the standard Charpy-V notch test due to fracture path deviation in to the softer base metal (Kalla & Funk, 1993).

The weld formation and quality of LB steel weldments are usually associated with three aspects: porosity, solidification cracking and high hardness in the HAZ and fusion zone.

a) Pores are formed as a result of dissolved gases or gases arising from contaminated surfaces, trapped process gases or evaporation of alloying elements.

Recent studies on different steel grades, thickness and welding speeds have showed that generally the porosity level associated with slower welds is higher than those connected with faster welding speeds. The evaluation of the weldment quality revealed that whilst higher porosity levels, irrespective of parent plate type, do not have a particularly detrimental effect on weld metal properties.

The high power density of the LBW process results in high processing speeds and therefore in short heating and cooling times (i.e. high heating and cooling rates). At increasing cooling rates the effect of alloying elements on the resulting microstructure decreases as martensite becomes the dominating structure. The maximum hardness of the martensite formed is determined by the amount of carbon in solution.

Laser beam welds in steels presenting solid state phase transformations are generally characterized by the presence of the hard and brittle structures in the fusion zone and HAZ which might have a negative effect on the deformability and toughness of the weldment.

- b) Solidification cracking is caused by segregation surrounding solidifying grains or dendrites of restrained welds.

In the presence of a minimum level of tensile stress or restraint, cracking occurs. In welding, the magnitude of these stresses is a function of joint geometry, particularly depth of weld penetration. It could be said that the susceptibility to form solidification flaws depends on: the geometry of the joint, the shrinkage involved in solidification, the coarseness of the solidifying structure and the amount and species of segregation. The narrow and deep geometry characteristic of laser welds results in very low bead width to bead height ratios.

The occurrence of solidification cracking has been shown to be directly related to external restraint. Recent studies have, however, revealed that small amounts of solidification cracks are non-critical to structures. Furthermore, it is possible to control or avoid these flaws by a combination of steel composition and welding parameter selection. Nevertheless, this may reduce the productivity of the laser welding process.

- c) Almost all LB welded structural C-Mn steels exhibit weld metal region of higher hardness and strength (possibly with lower toughness) compared to the base metal (overmatching) due to the rapid solidification and single pass natures of the welding process. The laser weld region with its high hardness and strength makes it almost impossible to determine the fracture toughness properties (intrinsic) of the weld region using conventional Charpy-V impact and Crack Tip Opening Displacement (CTOD) toughness testing specimens due to the crack path deviation towards the softer base metal (TWI Job Knowledge for Welders, 1997).

Various aspects of the similar and dissimilar joints of materials made by laser welding or other conventional material joining technologies still require a vast amount of research and development activities supported by the knowledge gained in long years of work on conventional materials. It is obvious that a better

understanding of the microstructure-mechanical properties relationship of laser welded joints will feed-back the materials development activities both in conventional and new materials.

2.1.8 Laser Beam Welding Applications

The laser beam welding is capable of joining a wide range of materials of interest in many industrial applications since it often offers remarkable advantages over conventional fusion welding processes. Thin sheet alloys for aerospace and automotive applications with minimal component distortion and high reproducibility of joint quality can be joined by laser beam welding. Body panels, transmission components and chassis members are laser welded on automobile production lines. Aerospace industry has used the unique properties of laser beam welding to incorporate materials previously considered difficult to weld, such as high temperature alloys, in their product designs, resulting in improved overall efficiency. The shipbuilding industry has a particular interest in the use of laser welding for joining large steel plates, provided certification by the classification societies is forthcoming. Another interesting aspect of the laser is the localized heating which makes it ideal for welding electronic components such as printed circuit boards, where high average temperatures even in small volumes surrounding the weld region is intolerable. More recent developments in laser beam welding include welding of heavier section steels for fabrication and construction. The most revolutionary application of laser beam welding is tailored blanks, which increase material utilization and improve product performance while reducing manufacturing and capital costs. Developments in process automation and on-line adaptive control will allow a wider range of industries to take advantage of the productivity, quality, flexibility and manufacturing opportunities provided by laser welding.

The application of laser welding will make an important impact on the whole concept of fabrication. The high accuracy to which the different elements can be welded, will have a high influence on the methods of assembly. The process will further make it possible to introduce new design concepts, which up to now have

been impossible to use due to the limitations connected with traditional welding techniques. Despite the high capital cost and greater demands to edge preparation, etc., it is obvious that laser welding in the coming years will break through in medium and heavy section welding industry.



2.2 Strength Mis-Match Effect in Welded Joints

Modern design and maintenance of engineering structures require quantitative and accurate assessment of the performance of welded or bonded components. Consequently, the assessment of the structural significance of strength mis-match, evaluation of the appropriate fracture toughness parameters and defect assessment procedures are attracting increasing interest.

A welded or bonded bi-material joint represents a common feature in many structural components. It is virtually impossible to produce absolute defect free welds and bi-material interfaces. Bi-material (dissimilar) joints often contain or develop defects in the form of interface cracks in the discontinuity area of the dissimilar compounds. The population and distribution of such defects associated with a joining process or service condition directly affects joint performance/life time of a structure by changing the strength and toughness properties. Mechanical properties of the mis-matched bi-material joint constituents will also influence the joint performance by changing the occurrence of a critical condition for unstable fracture. Mechanical properties of the mis-matched bi-material joint constituents (material A, interface structure or weld material and material B) will also influence the joint performance by changing the occurrence of a critical condition for unstable fracture. Since the weld metal and parent material are produced in totally different ways, often with different chemical analysis, heat treatment and microstructure, it is relatively common for them to have different mechanical properties. Differences in yield can have important effects on the way in which the joint behaves and on fracture toughness measurement and toughness requirements in service. The study of complex stress/strain fields around such interface cracks and the application of elastic-plastic fracture mechanics parameters (CTOD and J integral) under various loading modes are of great interest for the strength and fracture toughness analysis of mis-matched welded or bonded bi-material joints (Koçak & Schwalbe, 1994).

Fracture mechanics defect assessment procedures generally assume that any discontinuities found during fabrication or after a component has entered its service

life are located in material of uniform mechanical or microstructural properties. In practice, however, the different regions of conventional welded or bi-material joints exhibit significantly different mechanical properties. All fusion welding processes produce a heat affected zone (HAZ) which results in material properties different from those of base and weld deposit. Additionally, there are many structural components containing dissimilar (bi-material) weld joints with a high degree of strength mis-match and such joints are more susceptible regions for defects and/or cracks. The failure behavior of a welded structure associated with a defect in the weld zone will certainly be influenced by the differences of the strength levels of the weld metal (σ_{YW}), heat affected zone and base metal (σ_{YB}):

$$M = \frac{\sigma_{YW}}{\sigma_{YB}}$$

is the mis-match factor, i.e. the ratio of weld metal yield strength to base material yields strength; $M < 1$ characterizes undermatching, $M > 1$ refers to overmatching (Koçak, Saarenheimo, Petrovski, Talja & Schwalbe, 1994).

The effect of weld strength mis-match in service applications depends on the type of weld and its orientation relative to the applied stresses, and also on the dimensions of any crack like defects present relative to the width of the weld and to the width and thickness of the whole joint.

The subject of strength mis-matching effects on structural weldment's fracture characteristics and structural performance has recently received great attention world wide. A number of defect assessment methods have been developed which are supposed to cover the assessment of the severity of crack-like defects in welded joints. However, these methods are based on homogeneous materials and generally assume that defects present are located in material of uniform mechanical/microstructural properties and it is normal practice to use the tensile properties of the material in which the defect is located. However, in reality, the mechanical heterogeneity (differences in mechanical properties between the different

zones of the weld joint) will influence the plastic zone development process at defects and hence affect the relationship between crack driving force and applied loading (Koçak & Denys, 1994).

Determination of the tensile properties (degree of mis-match) and fracture toughness values of respective parts of the welds presents particular problems, due to the heterogeneous mechanical/microstructural nature of the joints and the small width of the HAZ as well as its microstructural complexity. Therefore, the use of existing standard test procedures for the mechanical testing of metallic materials requires some modifications on specimen preparation, testing and interpretation of the data for structural weldments. For example, Fig. 2.4 shows various mis-match cases of fatigue pre-cracked bend bars where either the weld metal or bi-material mis-match is considered (Koçak et al, 1994) .

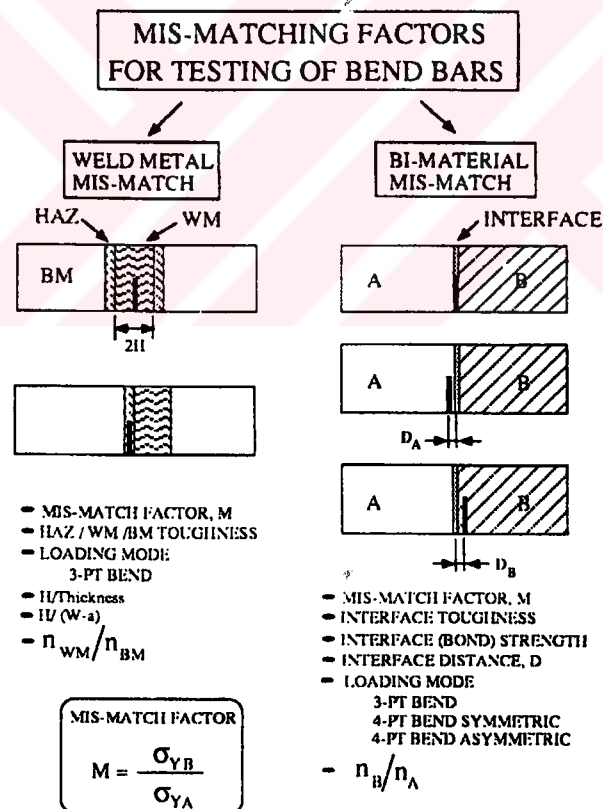


Fig. 2.4 Some factors to be considered in fracture toughness testing of mis-matched or bi-material bend specimen

There are no strength mis-match effects in the linear elastic regime if elastic modulus and Poisson's ratio are the same for the weld and base metals. In the plastic regime, strength mis-match influences the straining capacity, limit load analysis and hence fracture characteristics of welds by altering the plasticity development pattern (i.e., alteration of the proportion of remotely applied strain which reaches the crack tip) in the welded specimen from that characteristic of a cracked plain specimen. This redistribution of plasticity has significant influence on the wide plate and CTOD testing and on the implications of these tests with respect to the structural assessment of welds (Burdekin, Koçak, Schwalbe & Denys, 1993).

It is known that different parts of most welds exhibit significantly different tensile strengths and therefore, nominally identical applied strain levels cause different amounts of strain concentrations in the respective parts of the transverse weld joints, depending on the weld metal strength, crack size and geometry. The weld metal is usually being considered a potential site for defects or cracks to be present or develop in welded structures. Strength mis-matching often occurs as an overmatching weld in medium strength steel weldments. By doing this, a beneficial shielding effect on the higher yield strength weld metal (defective region in most cases) on a defect from imposed strains can be expected. When the weld metal yield strength is greater than that of the base metal (overmatching) plastic straining is likely to occur in the base metal (ϵ_{BM}), while the weld metal remains in the elastic regime (ϵ_{OM}) as shown schematically in Fig. 2.5.

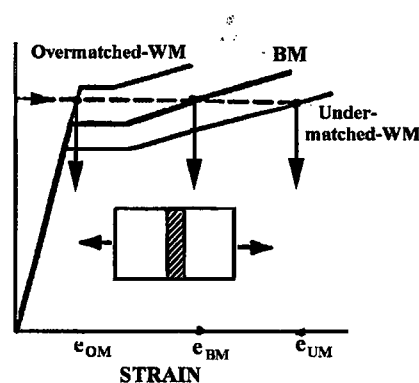


Fig. 2.5 Schematic s-e curves of a transverse weld joint showing development of different levels of strain on under or overmatched welds for a given stress

Plastic straining can occur in the base metal while the weld metal remains in the elastic regime. However, undermatched welds are sometimes used in high strength, low-alloy steel in order to avoid cold cracks, to obtain superior toughness due to lower strength, to increase the deposition rate and to reduce the residual stress thus reducing the fabrication cost by decreasing the required pre-heat. Lower weld metal yield strength than the base metal (undermatching) will cause a concentration of strain in the weld metal and local ductility of the undermatched weld metal will be exhausted prior to the onset of gross section yielding (GSY). In this case, the undermatched weld metal requires higher fracture toughness to prevent the risk of unstable fracture (Koçak & Denys, 1994).

It is rational to require a defective weldment to sustain remotely applied (gross) stresses of the base metal yield strength magnitude. Obviously, the requirement of base metal yielding for safe structural performance implies that the overmatching (yield strength of weld metal is higher than the base metal) case is desirable because yielding of the base metal (GSY) can be more easily obtained even for comparatively longer defects and furthermore, it implies that strain at failure will be plastic. However, the yield strength, weld width (2H) and strain hardening capacity of the weld metal region as well as defect position/shape/size will affect the yielding behavior of weldments. In particular, the interdependence of these variables should be established in order to determine the toughness needed for mis-matched weld materials for safety and economic considerations.

Various studies have addressed the fracture behavior of mis-matched weld joints and the results show that strength mis-match can significantly affect deformation and fracture behavior of the welded joints under bending and tension loading. Therefore, strength mis-match has a strong influence on both applied load and material toughness parameters based on the fracture mechanics parameters (CTOD or J integral) in bend and tension loaded specimens. The effect of the relative difference (mis-matching) of the yield strengths of the base, weld metal and HAZ parts on defect assessment procedures and on toughness values (CTOD and J) of the material

displaying a defect must be determined. There is a great need to establish a relationship between applied strain, toughness and defect size for mis-matched welds (Petrovski & Koçak, 1994).

2.2.1 Mis-Match in Laser Welded Specimens

It is common practice to obtain weld metals which display higher strength (overmatching) than the steels used in various engineering structures. By doing this, a beneficial shielding effect of the higher yield strength weld metal (defective region in most cases) on a defect from imposed strains can be expected.

Despite significant improvements in LB welding technologies over recent years, it still remains impossible to characterize narrow LB welds in a unique fashion and to produce quantifiable mechanical properties. General guidance on design aspects of LB welded joints is currently lacking for potential users so that they are often reluctant to opt for such a welding process despite the available technology to fabricate the joint. Present standards do not take into account the features characteristic of LB welds stemming from high energy density, high welding speed, high solidification rate and the small volume of base material affected. Many LB weld joints, therefore, fail according to the available standards, even though their mechanical and microstructural properties are good. Thus, additional non-standard tests, which consume time and money, are often requested by the industries.

There are no fracture mechanics based testing procedures available for power beam weld joints despite the wide and inevitable use in modern engineering structures. This discrepancy is due mainly to the lack of information on the interaction between the base metal and fusion zone, which have different mechanical properties. Substantial differences in strength properties (mis-matching) of the base metal and narrow fusion zone of the LB welds inevitably occur due to the rapid thermal cycle of the joining process. Considering the very narrow LB weld metal size, the “shielding effect” of the higher yield strength weld seam of C-Mn steels

(from externally applied strain) on a defect present within the weld seam could be a very different magnitude than the shielding effect observed on conventional larger sized weld joints.

Inevitable extensive plasticity development at the parent plate of the C-Mn steel laser welds may minimize the structural significance of the lower toughness small weld seam under some circumstances (i.e. depending on the temperature). If lower toughness weld or HAZ regions of the C-Mn steel welds cannot control the failure behavior of the standard laboratory specimen or component, no extra effort (i.e. artificial LB weld configurations or specimen design to force the crack to remain within the weld seam) should be made for the determination of fracture toughness of LB welds. Global deformation and fracture behavior of the LB weld joints which include inevitable participation of the neighboring base plate in the deformation process should be respected and considered. Therefore, the defect assessment procedures or standards developed for conventional weldments need to be modified for laser beam weldments. The fracture assessment of such welds for structural integrity requires detailed information on the overall fracture behavior. Presently, the lack of any international standard for incorporating mechanical heterogeneity (mismatch) in laser beam structural weldments and related safety assessment procedures urgently require the development of a technology applicable to civil, aerospace, oil and gas power generation, automotive, chemical, process and shipbuilding industries as well as being acceptable by designers and producers of structural materials (Koçak, Dos Santos & Çam, 1997).

For the time being, development of flat microtensile specimens (which will be discussed later in detail) which can be easily produced and tested in order to determine the mechanical property variations across the welded joints have been successfully utilized in laser beam welding, electron beam welding and diffusion bonding applications.

2.3 Austenitic and Ferritic Steels

Despite the development and influx of many new and advanced materials into the engineering world over the past 10 years, the family of stainless steels has persisted as preeminent in a wide variety of industrial systems that require exacting performance.

2.3.1 Austenitic Stainless Steel

Austenitic stainless steels typically have a composition within the range 16-26% chromium (Cr) and 8-22% nickel (Ni). A commonly used alloy for welded fabrications is Type 304 which contains approximately 18% Cr and 10% Ni. These alloys can be readily welded using any of the arc welding processes. As they are nonhardenable on cooling, they exhibit good toughness and there is no need for pre- or post-weld heat treatment (Job Knowledge for Welders, 1997).

Although austenitic stainless steel is readily welded, weld metal and HAZ cracking can occur. Weld metal solidification cracking is more likely in fully austenitic structures which are more crack sensitive than those containing a small amount of ferrite. The beneficial effect of ferrite has been attributed largely to its capacity to dissolve harmful impurities which would otherwise form low melting point segregates and interdendritic cracks (Lippold, 1992).

2.3.2 Ferritic Stainless Steel

Ferritic stainless steels have a Cr content typically within the range 11-28%. Commonly used alloys include the 430 grade, having 16-18% Cr and 407 grade having 10-12% Cr. As these alloys can be considered to be predominantly single phase and nonhardenable, they can be readily fusion welded. However, a coarse grained HAZ will have poor toughness (Job Knowledge for Welders, 1997).

The main problem when welding this type of stainless steel is its poor HAZ toughness. Excessive grain coarsening can lead to cracking in highly restrained joints and thick section material. In thicker material, it is necessary to employ a low heat input to minimize the width of the grain coarsened zone and an austenitic filler to produce a tougher weld metal. Although preheating will not reduce the grain size, it will reduce the HAZ cooling rate, maintain the weld metal above the ductile-brittle transition temperature and may reduce residual stresses. Preheat temperature should be within the range 50-250 °C depending on material composition (Dilthey & Risch, 1995).

Austenitic-ferritic dissimilar steel joints are widely used in power generation systems. For, example, in fossil fuel fired power plants, the primary boilers and heat exchangers operate at low enough temperature and under suitable environmental conditions, for economical ferritic steels to be chosen as construction materials. However, the high operating temperatures of the superheater and reheater require the use of more expensive austenitic stainless steels. Therefore, transition joints between the two types of steels are necessary. Such joints are normally produced using conventional welding processes. Investigations into the production of such types of joints and their failure analyses have been made by many researchers, and various measures have been used to improve the quality of the joints (Sun & Moisio, 1994).

2.3.3 Laser Beam Welding of Austenitic-Ferritic Steel Joints

In recent years, power beam technology - electron beam and laser beam - has developed significantly, and much attention has been paid to exploit its application in welding. The chemical composition (particularly carbon content) of the structural steels significantly influences the laser weldability of these materials. In modern structural steels the carbon content is significantly reduced and the strength is attained by alloying elements and/or thermal processing during rolling. These fine grained steel grains are particularly suitable for low heat input laser welding process to avoid the development of a coarse grained microstructure in the HAZ region.

However, laser beam welding, particularly autogenous welding, with its low heat input and high cooling rate (high welding speed), may present some drawbacks for producing austenitic-ferritic dissimilar steel joints through the formation of hard and brittle microstructures (i.e. martensite) within the narrow weld and HAZ regions of steels subjected to solid state phase transformations, which may result in a reduction in service properties. Since the hardness values reached in these regions are usually well above those specified in standards and codes, expensive and time consuming qualification procedures are required for individual components. Laser beam welding with filler metal is a relatively new technique introduced in order to widen the use of lasers in welding. Filler metal welding can provide some advantages over autogenous welding, in addition to retaining many of the features of laser beam welding, e. g., reducing the fit-up tolerance of laser beam welding; modifying the weld metal composition through the use of filler metal. Although laser beam welding of various dissimilar metals has been reported in recent years, little data for laser welding of austenitic-ferritic transition joints are available. As a new welding process for joining this type of dissimilar steel weld, whether it can be useful depends on how to take advantage of the process by selecting the correct processing parameters. Since laser beam welding can be operated in an exact and reproducible manner, the principles of selecting the optimum processing parameters becomes the most important knowledge to be obtained. There are a variety of processing parameters in laser beam welding such as: focusing conditions, shielding and plasma control conditions, also there are ones which specifically affect the quality of dissimilar steel joints such as root opening, beam alignment and heat input (Scrutton, 1995).

2.4 GKSS Research Center

The experimental part and the literature review of this study were carried out at GKSS Research Center. Therefore, a short introduction of this center is given in this section.

GKSS, with a staff of approximately 800 employees at its sites in Geesthacht near Hamburg and Teltow near Berlin, is one of the 16 national facilities belonging to the Herman von Helmholtz Association of German Research Centers (HGF).

GKSS has the task of opening up and shaping new strategic areas of benefit to the national economy, thereby establishing a basis for the key technologies of tomorrow.

About 85% of GKSS's ca. DM 120 million annual research budget is provided by the federal government (90%) and the state governments (10%). The contribution from additional earnings generated via contract research and licensing of GKSS patents for products and processes is increasing strongly.

An organizational structure based on a matrix system and project management principles serves to promote the interlinking of activities with partners from other research organizations, as well as private and public companies both nationally and internationally. Applications oriented work is based on fundamental research with clearly defined goals, supported by a sound scientific and technical infrastructure, which includes a research reactor, extensive experimental and testing facilities, mainframe computer and a research ship. Facilities for the production of membranes and advanced high temperature materials on a pilot-plant scale are examples of the bridges built to industrial applications.

2.4.1 Institute of Materials Research

GKSS work concentrates primarily on metallic materials for use in the fields of energy, technology and transport. The demands made on such materials are increasing rapidly. They must combine the advantages of metallic, ceramic and polymer materials while being easy to produce and process; and their behavior as component parts must be reliably determinable and predictable. From this starting point, the work done by GKSS moves into the spheres of the physics and mechanics of materials, also examining related methods of manufacturing and processing. The prime objectives of the work are:

- development of new high temperature lightweight materials (up to $> 800\text{ }^{\circ}\text{C}$ with 5 g/cm^3),
- explanation of mechanisms operating in the materials,
- comprehensive evaluation of the behavior of materials used in components.

Work in the field of materials development as such is therefore accompanied by process research geared to developing methods of manufacture, quality assurance, design of components and where necessary, even extending to pilot production. The requirements for manufacture, design and quality are specified at this stage in cooperation with the customers in the manufacturing, fabrication and utilization sectors of industry. Subsequent development work is undertaken as contract research, cooperative projects and basic research that is applications oriented.

Areas of expertise for contract research:

- Special, high strength, high temperature resistant alloys
- Fatigue behavior of components
- Formation of cracks in components
- International standards and test procedures
- Test procedures employing special test equipment

(Portrait GKSS 1006)

CHAPTER THREE

MATERIALS AND EXPERIMENTAL PROCEDURE

3.1 Materials and Experimental Procedure

In this study, austenitic stainless steel 1.4404 (X2 CrNiMo 18 10) and ferritic steels (grades St37 and St52) were used as base metals. Three similar (St52-St52, St37-St37 and 1.4404-1.4404) and two dissimilar (St52-1.4404 and St37-1.4404) single pass full penetration autogenous laser beam welds were produced without using filler wire at LAZ Nord, a laser cutting/welding firm in Hamburg using 15 kW cross flow CO₂ laser. Welds for each material were prepared from 6 mm thick, 235 mm x 200 mm plates clamped in flat position, the joint faces of which had previously been machined flat and cleaned using solvent. Similar and dissimilar welded joints are schematically shown in Fig. 3.1.

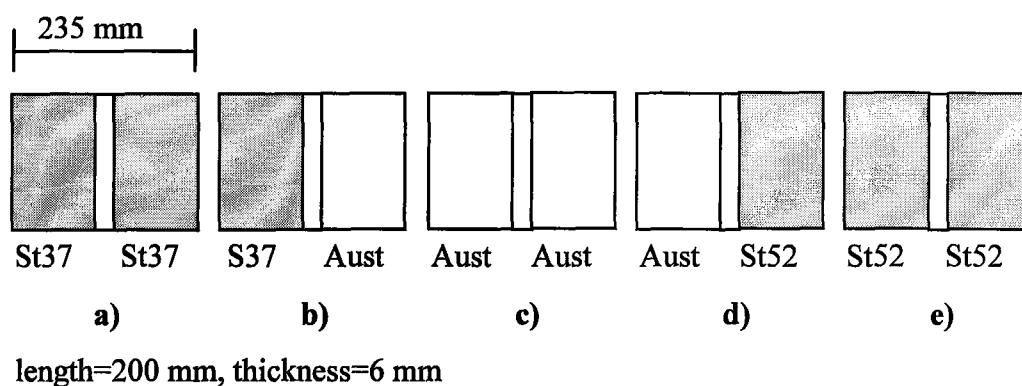


Fig. 3.1 Schematic illustration of: a) Similar St37-St37, b) Dissimilar St37-Aust, c) Similar Aust-Aust, d) Dissimilar St52-Aust, e) Similar St52-St52 joints

Each weld was visually inspected to ensure that its fusion position and external profile were satisfactory, and the presence of porosity and other weld defects such as centerline cracking and lack of penetration were assessed by X-ray radiography. Extensive microhardness measurements (using 100 g. load) were performed across the welded regions at three different locations, namely at the weld root, mid-section and top part of the joints. In addition, the HAZ regions of the ferritic steels were extensively screened by conducting microhardness measurements parallel to the fusion line.

Standard round tensile specimens of the base metals were tested in order to determine the base metal mechanical properties and mis-match ratios. Standard flat transverse tensile specimens were extracted from the welded plates and tested at room temperature, Fig. 3.2. Weld reinforcement was machined before testing. Photoelastic specimens were prepared in order to see the deformation patterns when the specimen is subjected to tensile loading.

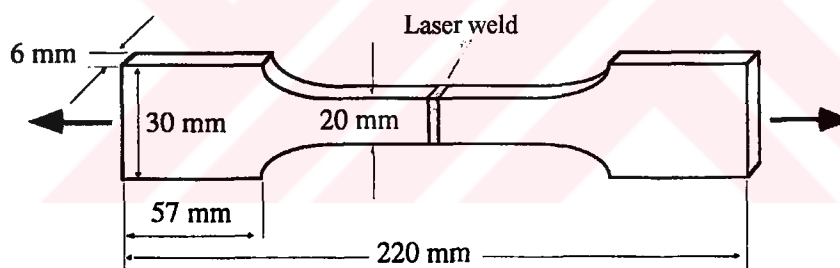


Fig. 3.2 Transverse tensile specimen

Sets of flat microtensile specimens were extracted by spark erosion cutting from base metals, HAZ's and weld metals of all the joints studied. Preparation of flat microtensile specimens by spark erosion cutting technique is described in Section 3.2. All tensile tests were carried out at room temperature using a displacement rate of 0.5 mm/min in a screw driven universal testing machine. The load and extension data were continuously recorded by a data acquisition system during the tests. After testing, the broken half of the specimens were mounted for microstructural

verification of the specimen location. The fracture surfaces of selected specimens were also examined by scanning electron microscopy. Microhardness measurements on the non-deformed specimen heads were also carried out.

Single Edge Notch Bend (SENB) specimens which have a thickness of 6 mm were prepared from the base metal, HAZ and weld regions in order to measure the Crack Tip Opening Displacement (CTOD) and Crack Mouth Opening Displacement (CMOD) values. SENB specimens, which were machine notched in the weld zone, were further fatigue pre-cracked in order to introduce a sharp crack ($a/W=12.5/25=0.5$) from an initial of 9 mm crack length to 12.5 mm, as described in Section 3.3. CTOD measurements obtained by three point bending tests, using the GKSS developed δ_5 clip gauges were conducted on SENB specimens which allowed us to make a direct measurement of the crack tip opening displacement for each specimen, this technique is described in Section 3.4. Again photoelastic specimens were prepared to obtain the deformation patterns occurring during loading.

Optical and scanning electron microscopy were conducted before and after testing to determine the details of the fracture process at the weld region. For optical metallography, specimens were mounted in bakelite and ground with silicon carbide papers of 1200 and 2400 grades followed by polishing on a rotating wheel with 6, 3 and 1 micron diamond paste and a final polishing with OPS cloth. Ferritic steels were etched with a solution comprising 2% nitric acid in alcohol for 10 seconds, while austenitic steels were etched at 70 °C with V₂A-etchant, which consists of 100 ml distilled water, 100 ml HCl, 10 ml HNO₃ and 0.3 ml Vogels Sparbeize (etching solution), for 5-10 seconds.

Stress-strain curves, hardness vs. distance from the weld center, crack tip opening displacement-load, crack mouth opening displacement-load, CTOD vs. hardness and other comparison diagrams and photographs of macrosections, micrographs and fracture surfaces are prepared from the results of the tests conducted.

3.2 Preparation of Flat Microtensile Specimens

“Flat microtensile specimen” testing procedure allows the determination of the tensile property gradient existing in the LB fusion zone. The flat microtensile specimen technique has originally been developed for property determination of HAZ of conventional multi-pass weld joints. Successful applications of this technique for thick section similar and dissimilar electron beam welds (Koçak & Junghans, 1994) and for the bond strength determination of the diffusion bonded joints has been carried out at GKSS Research Center.

The extraction of flat microtensile specimens in order to obtain the local mechanical properties at the weld region, HAZ and base metal was carried out in two stages:

- i) extraction of a pre-shaped block with laser weld in the middle
- ii) cutting out specimens from etched pre-shaped block by spark erosion cutting technique (using 0.1 mm diameter Cu-wire) parallel to the weld, as schematically shown in Fig. 3.3.

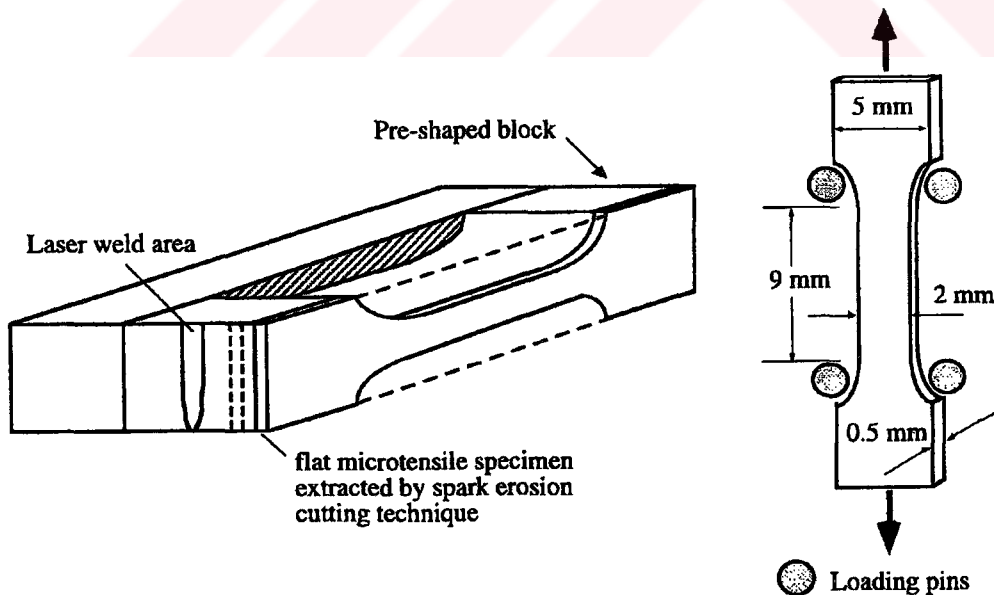


Fig. 3.3 Cutting out flat microtensile specimens from etched pre-shaped block

Comparison of a flat microtensile specimen's dimensions to a paper clip is shown in Fig. 3.4. Due to the small size of the microtensile specimens, loading was introduced using four high strength round pins at the shoulders of the specimens, Fig 3.5 (Çam, Riekehr & Koçak, 1997).

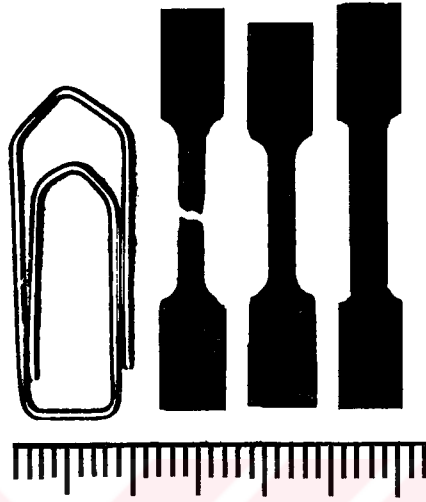


Fig 3.4 Comparison of a flat microtensile specimen's dimensions to a paper clip



Fig 3.5 Loading of the specimens using four high strength round pins at the shoulders

3.3 Fatigue Pre-Cracking

The present fracture toughness testing procedures require fatigue pre-cracked specimens with similar restrictions on the fatigue crack front shape, length and the loading conditions (particularly with regard to the level of maximum load and stress ratio, R to be used). To meet such requirements in weldments, a special pre-cracking procedure should be used to obtain uniform fatigue pre-crack growth across the weld thickness for specimen blanks containing welding residual stresses. The through thickness pattern of the welding residual stresses (transverse to the weld length) changes from tensile (+) stresses near the surfaces to a compressive (-) ones as balancing stress at about mid-thickness. These residual stress components act as additive stresses to the applied stresses during the cyclic loading. Therefore, during fatigue pre-cracking, compressive stresses (at mid-thickness region) normal to the crack plane can counteract to the applied cyclic stresses and thereby decrease the magnitude of effective stress intensity range, thus inhibiting crack growth there.

Meeting the requirements in the CTOD testing of welds requires the use of modified pre-cracking techniques such as High- R ratio, Reverse Bending and Local Compression techniques. The modified version of the High R -ratio method as a “Step Wise High R -ratio (SHR)” was proposed by Koçak et al. With the help of SHR technique, improvements on the fatigue crack shape of the as welded CTOD specimens can be achieved. This technique uses an allowable $F_{\max}$ value and simply consists of two R -ratio levels. The basic principles of this technique are schematically shown in Figure 3.6 (Schwalbe & Koçak, 1992).

Step I($R=0.1$)= The first step can be used to initiate and propagate the fatigue crack to the length of about 1.0 mm. The initiation period of the fatigue crack from a machined notch will be minimized by using the full range of the applied load. During this step, a minimum or no crack growth will occur at mid-thickness of the specimen.

Step II (R=0.7)— In this step, the R-ratio of the cyclic loading is simply increased to 0.7 by keeping the same $F_{\max}$ as in the first step. The R-ratio should be used to see the improving effect on the crack front shape and to propagate the fatigue crack to the required length (Fig. 3.6). The use of R=0.7 from the beginning of pre-cracking will increase the time of pre-cracking considerably. Therefore, it is recommended to use R=0.1 only for the initiation stage of the fatigue crack at the machine notch tip in order to minimize the total pre-cracking time. It should be noted that decrease of the applied stress amplitude in Step II will obviously increase the time for completion of fatigue pre-cracking.

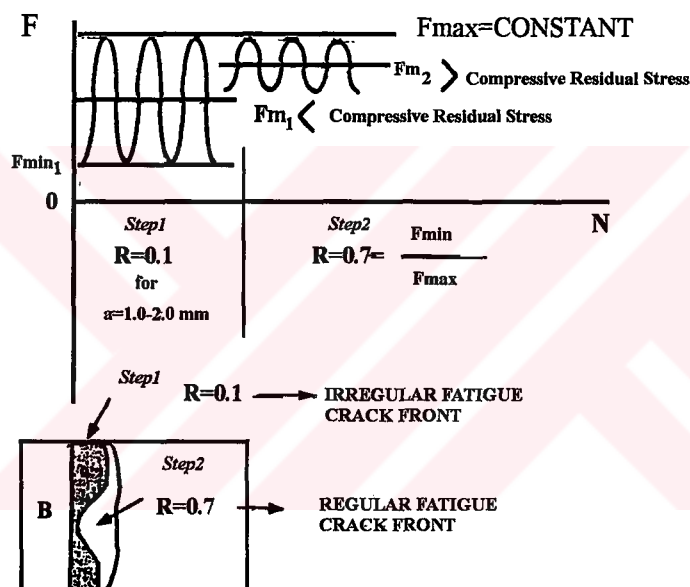


Fig. 3.6 Schematic diagram showing the Stepwise High R-Ratio fatigue pre-cracking method

The increased level of mean load $F_m = (F_{\max} + F_{\min})/2$, for a given $F_{\max}$ will, in fact, prevent a possible crack tip closure at the compressive residual stress region of the specimen. The applied static load of F_m should be high enough to keep the crack tip open at about midthickness region of the specimen by balancing the compressive stresses. Experience has shown that in most cases, using the R-ratio of 0.5 may not

provide a high enough mean load to prevent the retardation of the crack growth at the compressive residual stress region.

A fatigue pre-crack must be produced in the test specimen from the end of the machined notch, Fig. 3.7, to give an initial crack length a_0 in the range $0.45 \leq a_0/W \leq 0.65$ for SENB specimens (Schwalbe, Neale & Heerens, 1994).

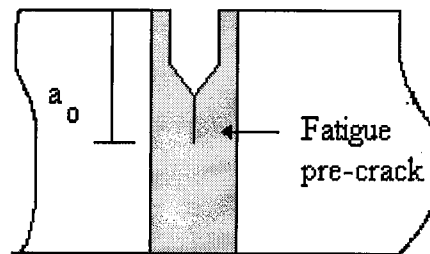


Fig. 3.7 Fatigue pre-cracking in a SENB specimen

The yield force for a SENB specimen is given as

$$F_y = \frac{4B(W-a)^2}{3S} R_{P0.2} \quad (\text{kN})$$

where B is the thickness of the specimen (mm), W is the width (mm), a is the initial crack length (mm) and $S=4W$ is the span between the axes of the two bottom rollers of the fatigue pre-cracking machine. The maximum fatigue force must not exceed the lower of $0.6F_y$. For example when $R=0.1$, the conditions below are taken for austenite:

$B=6 \text{ mm}$

$W=25 \text{ mm}$

$a=9 \text{ mm}$

$S=4W=100 \text{ mm}$

$R_{p0.2}=308 \text{ MPa}$

The following Table, to be used in fatigue pre-cracking tests is obtained.

Table 3.1 Values for austenitic SENB specimen used in fatigue pre-cracking tests

	F_y (kN)	X	$F_{max}=X \cdot F_y$	F_{min}	F_{stat}
B=6 mm W=25 mm a=9 mm R=0.1 $R_{p0.2}=308$ MPa	6.31	0.10	0.63	0.06	0.35
	6.31	0.11	0.69	0.07	0.38
	6.31	0.15	0.95	0.09	0.52
	6.31	0.20	1.26	0.13	0.69
	6.31	0.23	1.45	0.15	0.80
	6.31	0.25	1.58	0.16	0.87
	6.31	0.30	1.89	0.19	1.04
	6.31	0.35	2.21	0.22	1.21
	6.31	0.40	2.52	0.25	1.39
	6.31	0.45	2.84	0.28	1.56
	6.31	0.50	3.16	0.32	1.74
	6.31	0.55	3.47	0.35	1.91
	6.31	0.60	3.79	0.38	2.08
	6.31	0.65	4.10	0.41	2.26
	6.31	0.70	4.42	0.44	2.43

It is obvious that this technique can easily be applied to any specimen geometry and weld type on any standard testing machine without any extra operation and set-up. This is an important simplification on valid specimen preparation for the fracture mechanics testing of the weldments in any section size.

3.4 Fracture Mechanics Tests

The crack tip opening displacement (CTOD) is a measure of the stresses and strains near the crack tip and can be used to characterize the material's fracture resistance and although the severity of applied strain in a cracked structural element can be expressed in various ways (by remote, gauge length or local strains), the use of local strain in defect assessment procedures could be more realistic in characterizing the strain imposed on a structural element containing a weld since failure of such elements/welds is a strain controlled event governed by local strain.

However, the present defect assessment methods which have been developed to cover the assessment of the severity of crack-like defects in welded joints are based on homogeneous materials and generally assume that defects present are located in material of uniform mechanical/microstructural properties and it is normal practice to use the tensile properties of materials in which the defect is located. In reality, the mechanical heterogeneity will influence the plastic zone development process and stress state at the crack tip and hence affect the relationship between crack driving force and applied loading and thus the obtained standardized fracture toughness parameters (CTOD or J). The problem of measuring a meaningful fracture toughness value (in terms of standardized CTOD or J) of mis-matched laser weld specimens is very hard to overcome due to the high strength mis-match within very small region at the vicinity of the crack tip. It is not a simple task to distinguish the contributions from both parent materials (lower strength) and narrow LB weld (highly overmatched) at the vicinity of the crack tip to the remotely measured crack mouth opening displacement (CMOD) and load line displacement (V_{LL}) usually used in standardized CTOD and J estimates. The use of present standards for the testing of welds and bi-material joints requires significant modifications on specimen preparation, testing and evaluation of the CTOD (Koçak & Schwalbe, 1994).

In order to overcome these difficulties a local and direct measurement technique (δ_5 technique) has been developed at GKSS Research Center as a measure of the CTOD for determining the fracture toughness and the crack growth resistance. It

consists of measuring the relative displacement of two relative gauge points directly at the crack tip using special displacement gauges. The resulting CTOD is called δ_5 because the gauge length over which the CTOD is determined amounts to 5 mm. The advantage of this measurement concept is that the δ_5 type CTOD can be easily measured on any configuration with a surface breaking crack; no calibration functions are required. A further appealing property of the δ_5 technique is that since it is measured locally as a displacement at the location of interest, it does not have to be inferred from remotely measured quantities, like the J integral or the standardized CTOD. This is of particular importance when the specimen is mechanically inhomogeneous, as is the case for highly strength mis-matched narrow laser welds and in this case the δ_5 clip permits an easy measurement of the CTOD which can be considered as a crack driving force. The CTOD values of the laser weld joints on the SENB specimens have, therefore, been measured in terms of the CTOD (δ_5) technique (Petrovski & Koçak, 1994).

δ_5 , as a parameter characterizing the crack tip behavior, specifically has the following advantages:

- δ_5 is a quantity which is locally measured; the measurement is independent of the global behavior of the specimen.
- δ_5 is directly measured as a displacement; no calibration function is needed.
- It is of particular interest for mis-match welds and for interface cracks.

For the application of the δ_5 clip technique the area around the expected fatigue crack propagation path should be polished. After fatigue pre-cracking, Vickers hardness indentations are placed ± 2.5 mm on either side of the crack tip normal to the tip position, to give a gauge length of 5 mm. A δ_5 clip gauge with needle tips is seated into the hardness indentations and attached to the specimen using a lever mechanism. This is shown for a Compact tension (CT) specimen in Fig 3.8.

The basic arrangement for the measurement of the CTOD by δ_5 clip gauges is shown in Fig. 3.10 for a SENB specimen. Two δ_5 clips can be used for front and back faces of the specimen. CMOD measurement is accomplished by placing a clip, Fig. 3.11, at the mouth of the notch (GKSS Displacement Gauge Systems for Applications in Fracture Mechanics, 1995). The test is terminated when the load starts to decrease after reaching the maximum load or in the case of St52-St52 tests at $-40\text{ }^{\circ}\text{C}$ after the pop-in. The CTOD value is measured at the maximum load or in the case of $-40\text{ }^{\circ}\text{C}$ tests, it is measured either at the point of pop-in or maximum load.

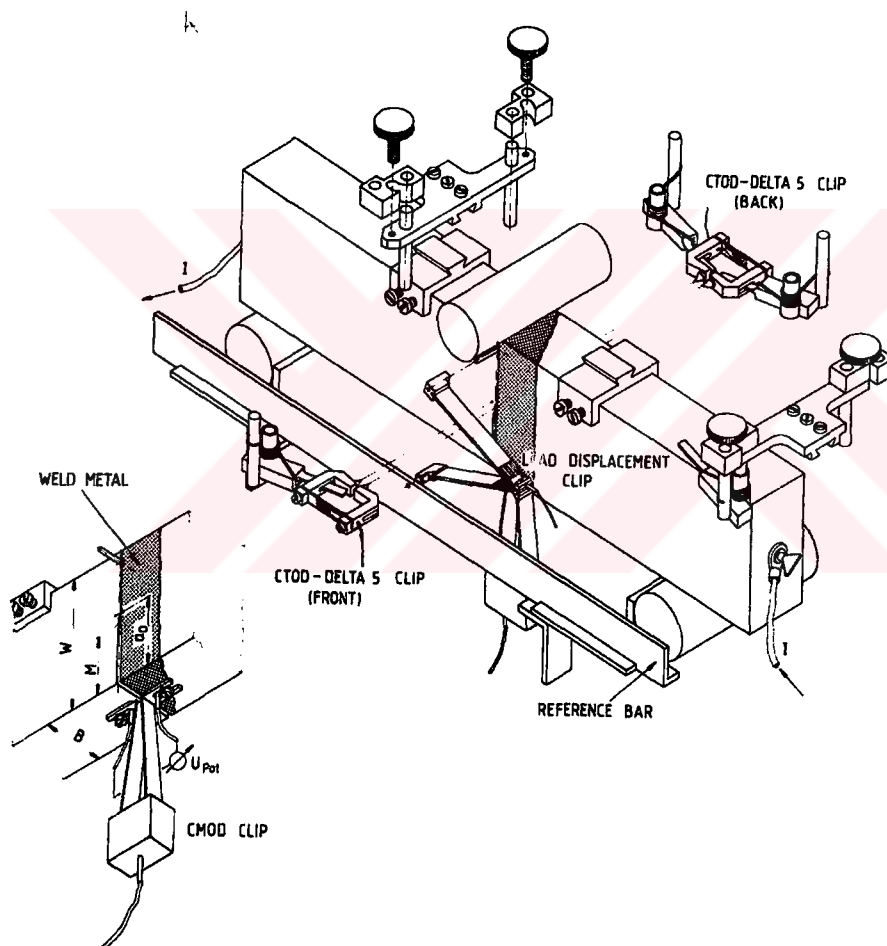


Fig. 3.10 Instrumentation of fracture toughness specimen with δ_5 -clips for HAZ testing

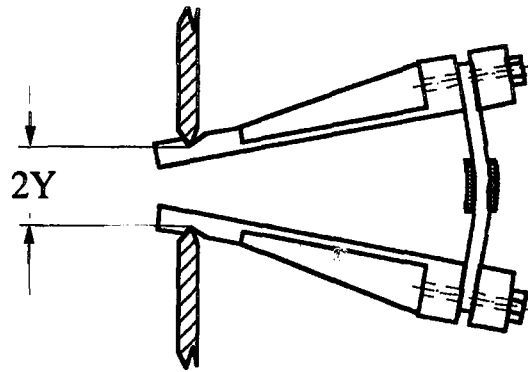


Fig. 3.11 CMOD gauge mounting

Three point bending tests at $-40\text{ }^{\circ}\text{C}$ are conducted by placing a chamber around the testing fixture and pumping nitrogen gas into this chamber, thus decreasing the temperature inside the chamber to the desired test temperature.

Pop-in Behavior: In Fig. 3.12, an example of a force-displacement test record containing a pop-in for the St37-St37 tests at $-40\text{ }^{\circ}\text{C}$ is shown. The pop-in step is usually characterized by an abrupt force drop and a small increase in displacement. The behavior is associated with the arrest of a rapidly moving crack during unstable crack growth which often appears as a distinct brittle thumbnailed region on the fracture surface of a specimen (Schwalbe, Neale & Heerens, 1994).

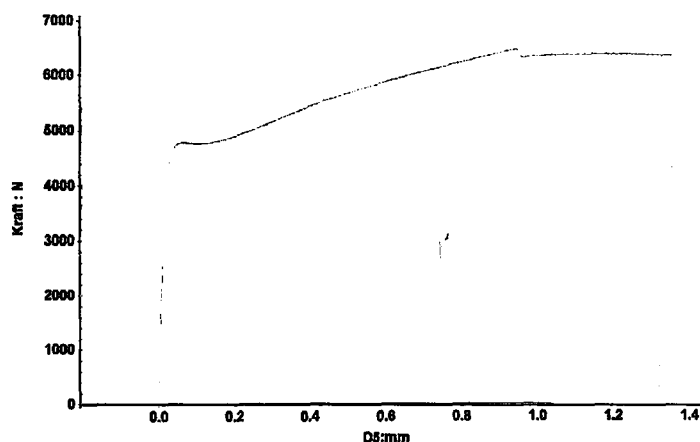


Fig. 3.12 Example of a pop-in behavior in $-40\text{ }^{\circ}\text{C}$ St37-St37 CTOD Measurement

The δ_5 testing technique permits direct measurement of the CTOD, of which the use is not limited and rather universal and it offers a unique capacity to assess the fracture toughness of bi-material interfaces by not taking into account of any remote plasticity. It can be applied to any type of testing pieces as well as for structural components with surface breaking cracks.



CHAPTER FOUR

RESULTS AND DISCUSSION

4.1 Microstructure

4.1.1 Austenitic Steels

The avoidance of solidification cracking is the most important requirement in the fusion welding of austenitic stainless steels. The deep and very narrow weld beads and rapid solidification rates of laser welding process may lead to solidification cracking although this steel can be crack-free welded by arc welding process. It has been reported that differences exist between the solidification behavior and the microstructure of austenitic welds experiencing the more moderate cooling rate involved in conventional welding processes and the much faster cooling rates characteristic of the high energy density welding processes like laser and electron beam welding (Koçak & Junghans, 1994).

During solidification, low melting segregates are entrapped along cell boundaries as spherical particles. Because of the lack of large-angle boundaries, these droplets usually cannot form films and propagate cracks. For 100% primary austenite solidification, however, rather flat, smooth crack paths exist along the grain boundary. With the presence of low-melting segregated liquids and welding stresses, cracks easily form and propagate as long as the liquid phase wets the boundary. Solidification cracking results from the segregation of low melting liquids along the grain boundaries during the last stages of solidification. If sufficient stresses are generated before final solidification, the boundaries may separate and form cracks. Primary austenite solidification becomes predominant with decreasing Cr/Ni ratio. The fraction of austenitic solidification also increases with the weld speed, a result which might be expected since a rapid cooling favors the formation of metastable

austenite (Brooks, 1992). Welds that solidify as primary ferrite are much more resistant to solidification cracking than welds that solidify as primary ferrite.

4.1.2 Ferritic Steels

Modern steels often contain relatively low amounts of carbon and only sufficient amounts of other elements influencing the hardenability, primarily manganese. The high cooling rates required to produce martensite in such steels are not usually obtained in arc welding processes and only seldom will the brittle element martensite be present in large amounts. This situation may dramatically change in high speed autogenous welding of the same steels. The thermal cycle is sufficiently fast that martensite often forms in laser welding of structural C-Mn steels. Depending on the actual thermal cycle, laser weld microstructures thus may contain appreciable amounts of martensite and may even be fully martensitic. The formation of martensite may result in a reduction in service properties, since martensite is a hard, low ductility phase and should generally be avoided in structural applications (Kristensen & Borggreen, 1994).

4.1.3 Dissimilar Joints

The microstructure of dissimilar weld metal can be fairly inhomogeneous. Depending on the degree of dilution from each side, various mixtures of austenite, ferrite and martensite phases could be found, ranging from a mainly austenitic to a predominantly martensitic structure. Regions of the weld zone microstructure poor in Cr and Ni are predominantly martensitic, while regions rich in the same elements can exhibit a mainly austenitic-ferritic structure. The narrow HAZ formation on the ferritic steel side can distinctly be observed contrary to the austenitic steel side of dissimilar laser weld joints where there is no distinct HAZ.

4.1.4 Microstructural Examinations of Similar Joints

Microstructural examinations of the joints investigated showed that the weld regions of all the joints containing ferritic constituent contained bainite and martensite, which is a hard and low ductility phase produced due to the high cooling rate in LB welding. Ferritic similar joint St37-St37 displayed a weld metal structure consisting of grain boundary ferrite, martensite and bainite, Fig. 4.1 and the other ferritic similar joint, St52-St52 contained martensite and bainite, Fig. 4.2. Austenitic similar weld, 1.4404-1.4404, exhibited an austenitic-dendritic (cellular) structure with no evidence of martensitic formation as expected, Fig. 4.3. Solidification cracking has not been observed in similar austenitic joints which might be due to a very fine austenite cell size characteristic of primary austenite solidification involved in laser beam welding of this steel grade. Contrary to the austenitic steel weld joint where no distinct HAZ is observed, A distinct HAZ development at the similar ferritic joints was observed. The HAZ in these joints contained a refined ferrite/pearlite structure and pearlite dissolution at the base metal sides.

4.1.5 Microstructural Examinations of Dissimilar Joints

Ferritic-austenitic dissimilar joints, on the other hand, showed an inhomogeneous weld metal microstructure, changing from the ferritic to the austenitic side, Figs. 4.4 and 4.5. A narrow HAZ on the ferritic steel sides was observed contrary to the austenitic side where there is no distinct HAZ. This is typical of dissimilar laser steel weldments. The HAZ on the ferritic steel side of the St37-austenitic joint, Fig. 4.4, contains no martensite in contrast to that of the St52-austenitic joint, Fig. 4.5, which exhibits a martensitic structure. The rest of the weld in the middle displayed an inhomogeneous microstructure which contained ferritic-austenitic and bainitic and/or martensitic phases in varying degrees due to an incomplete mixture of the molten metals of both sides in the weld pool during solidification. Some austenitic-dendritic (cellular) structure was also observed at the austenitic side. Moreover, some solidification cracks were also observed in the weld zone of the dissimilar joints, Fig.

4.6, which indicates that sufficient stress levels were developed in these weldments separating the grain boundaries in the cellular structure (presenting rather flat smooth crack paths) during solidification. It can also be assumed that low-melting constituents have been segregated along these boundaries, which resulted in solidification cracking during the last stages of solidification.

Microstructural observations and microhardness measurements on the gauge length regions of the microtensile specimens were also carried out for microstructural verification of the specimen location and to correlate the mechanical properties to microstructure. It is essential to understand the relationship between microstructure and mechanical properties with optimum strength and toughness of weld metals. Therefore, the interaction of the different microstructural constituents is especially important in terms of their effect on toughness. These microstructural observations will be discussed in Chapter 4.3.

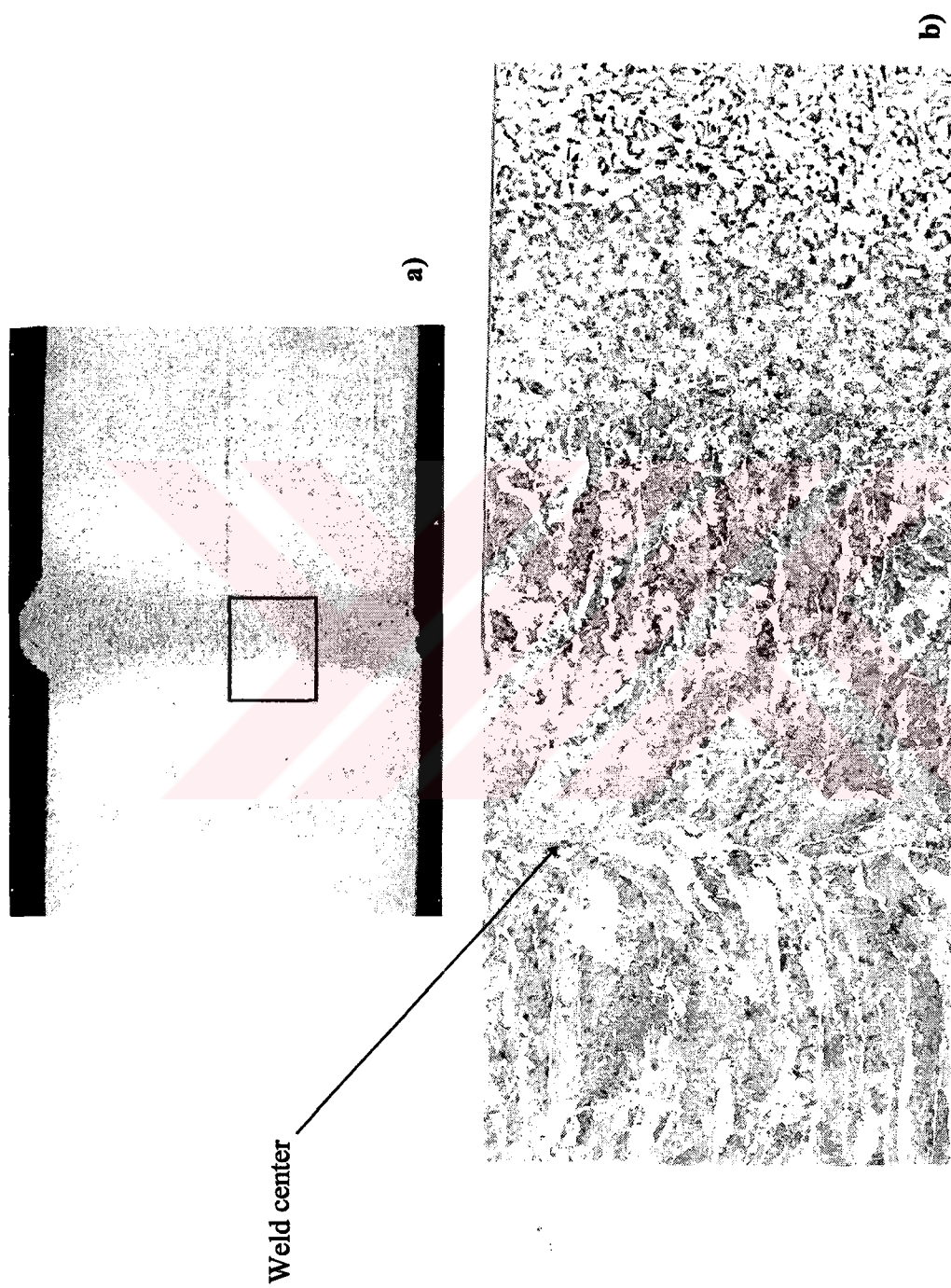


Fig. 4.1 Similar St37 LB joint a) macrosection, b) higher magnification of the region indicated in a)

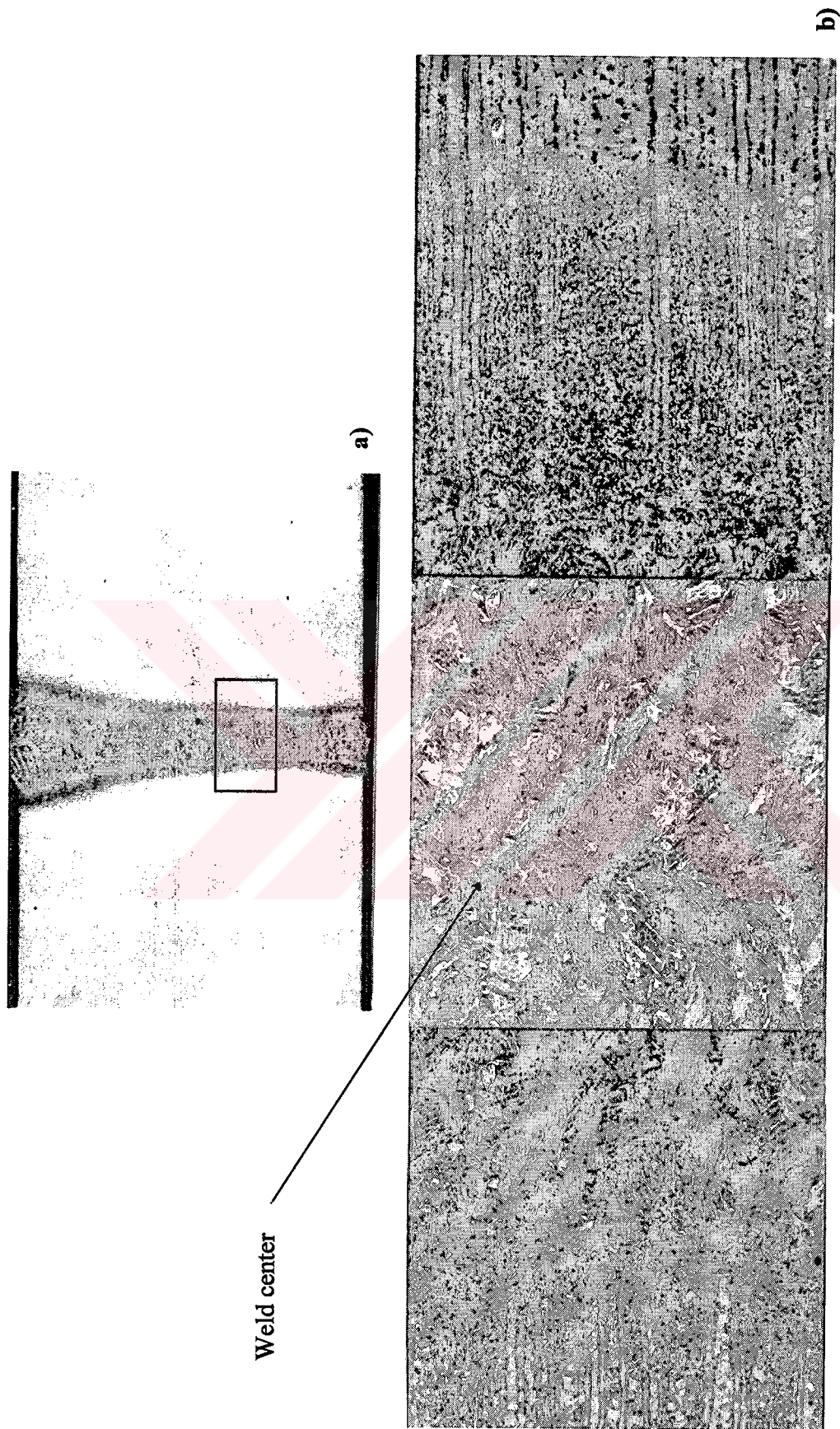


Fig. 4.2 Similar St52 LB joint a) macrosection, b) higher magnification of the region indicated in a)

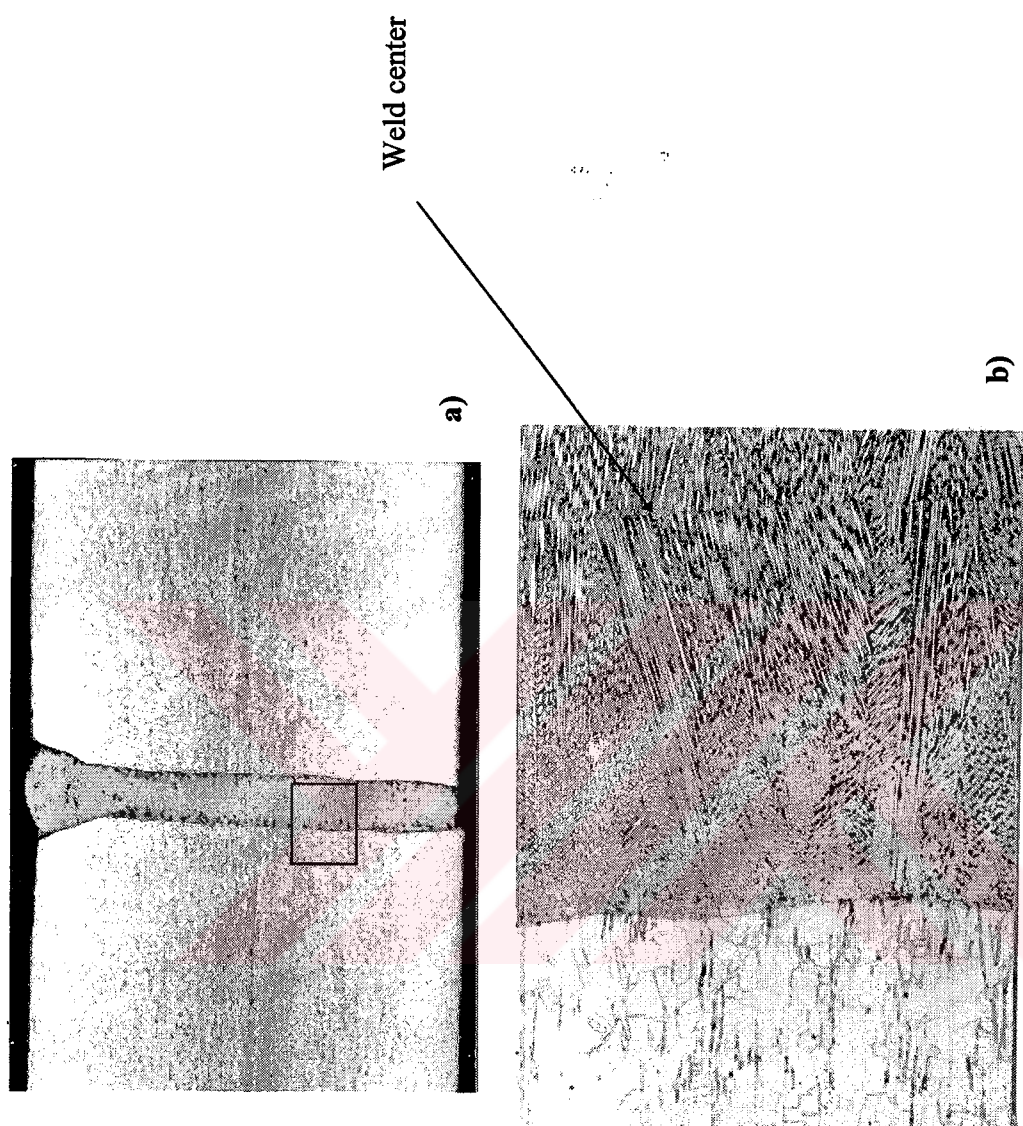
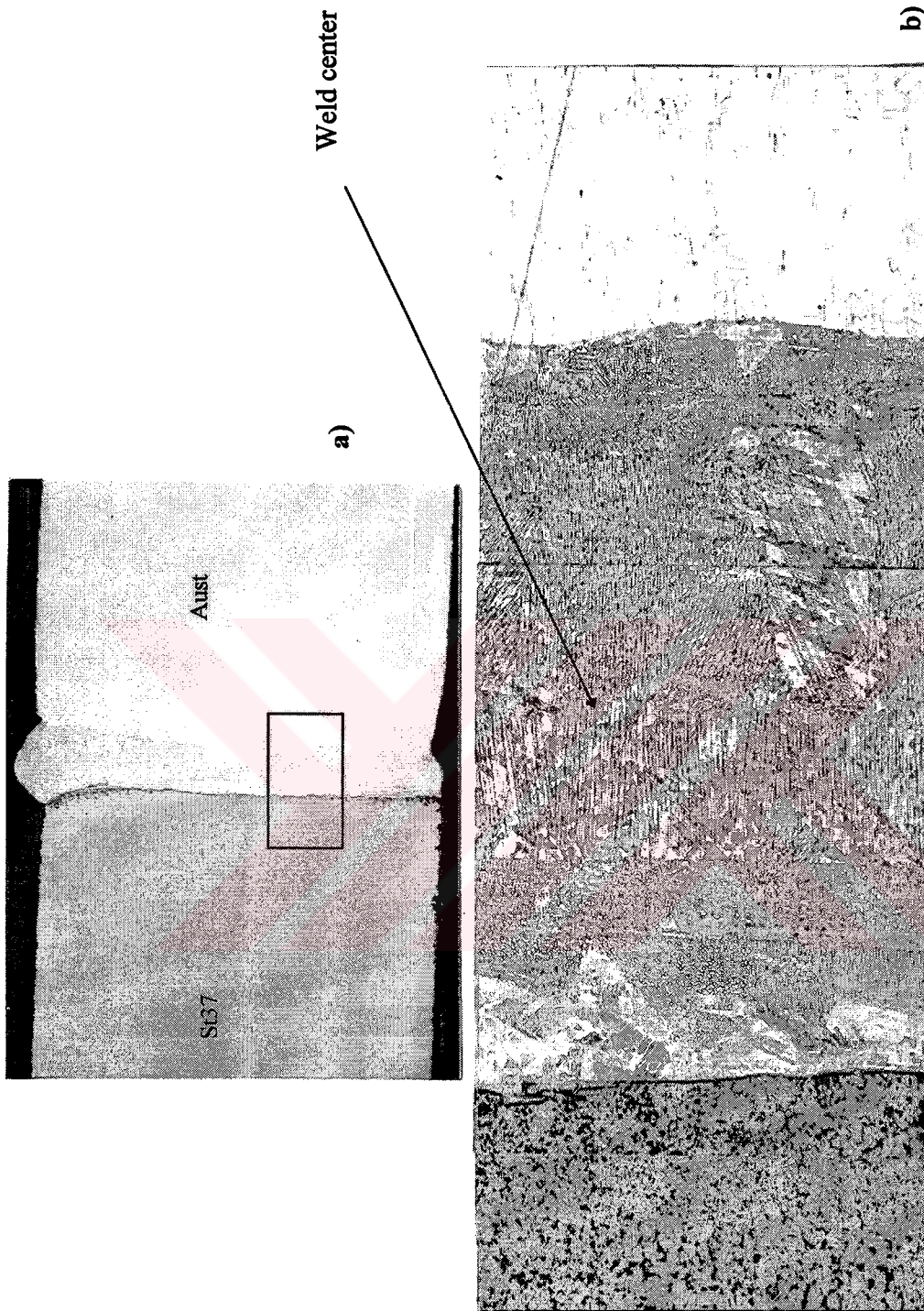


Fig. 4.3 Similar aust LB joint **a)** macrosection, **b)** higher magnification of the region indicated in **a)**



Dissimilar St37-aust LB joint a) macrosection, b) higher magnification of the region indicated in a)

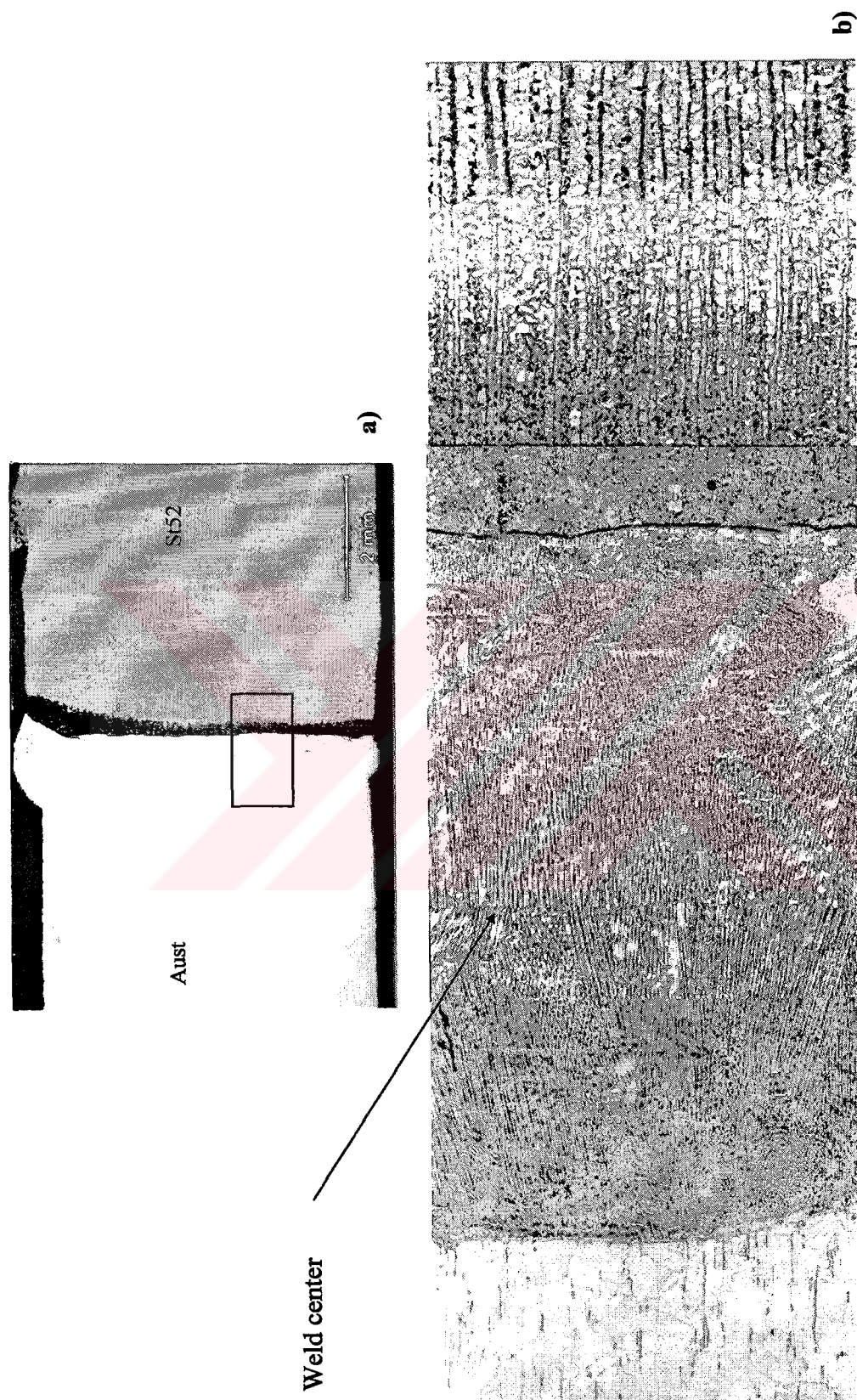


Fig. 4.5 Dissimilar St52-aust LB joint a) macrosection, b) higher magnification of the region indicated in a)



Fig. 4.6 Solidification cracking in the weld zone of dissimilar joint

4.2 Hardness

The welding process may lead to drastic changes in the microstructure with accompanying effects on the properties of the joint. The thermal cycles associated with laser welding are generally faster than those involved in the conventional welding processes. This is a natural consequence of the low heat input and high welding speed (rapid cooling rate). It does, however, mean that the weld zone will locally exhibit a high hardness for C-Mn structural steels. In order to determine the hardness profiles, extensive microhardness measurements were conducted at three locations (top, middle and root) as shown in Fig.4.7. Vickers method was utilized which is suitable for narrow laser welds since the measurements can be performed at a range of indentation forces. The set-up used for measuring the hardness profiles is seen in Fig. 4.8. The microhardness results exhibit no difference between these three locations for all the welded joints studied. It can, therefore, be concluded that no significant gradient in mechanical properties of laser welds along plate thickness direction is present.

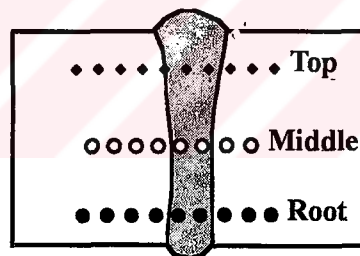


Fig. 4.7 Schematic illustration of the microhardness measurement procedure

Similar ferritic joints displayed a high hardness profile in the weld region while austenitic weld showed no significant increase across the joint. Similar St37 and St52 joints exhibited hardness values of 270-350 HV (Fig. 4.9) and 405-450 HV (Fig. 4.10), respectively, whereas similar austenitic joints displayed hardness values of about 210 HV in the weld region, Fig. 4.11. This is expected due to the lack of bainite and/or martensite formations in the weld regions of austenitic joints.

Dissimilar St37-austenitic joints displayed a hardness peak in the weld metal which exhibited hardness values of about 380-400 HV, Fig. 4.12. On the other hand, dissimilar St52-austenitic joints, Fig. 4.13, displayed a hardness peak of about 425 HV at the ferritic HAZ side slightly higher than that of the weld metal which exhibited hardness values of about 400 HV, which is due to the presence of martensite in the narrow HAZ region at the ferritic side as mentioned earlier. High hardness values obtained in the weld regions of ferritic similar and dissimilar joints are due to the presence of bainite and/or martensite in the microstructure. The presence of high hardness peaks indicate strength overmatching nature of the weld joints.

Extensive microhardness measurements were also conducted on the non-deformed heads of the microtensile specimens after testing. St37 base metal in the case of similar joint exhibited an average hardness value of 128 HV, whereas that of dissimilar joint displayed a hardness of 147 HV. For St52 base metal the values for similar and dissimilar joints are 175 HV and 190 HV, respectively. This might be due to the difference in welding direction of these sheets with respect to the rolling direction. Austenitic base metal showed an average of 175 HV. Microtensile specimens extracted from the weld metals of similar ferritic and dissimilar joints exhibited a significant increase in hardness, average hardness values being 347 HV and 404 HV, respectively for St37 and 450 HV and 430 HV, respectively for St52 which is due to the formation of bainite and/or martensite whereas the weld metal of austenitic similar joint did not exhibit a significant hardness peak, the average hardness value being 207 HV. The HAZ of the St37 similar joint displayed hardness values in the range of 138-165 HV, the average value being 152 HV. An abrupt change is observed in the hardness values of the HAZ region of the St52 joint. In this case, the range is between 190-222 HV for the similar joint. The HAZ's at the ferritic sides of the dissimilar joints showed hardness values in the range of 169-216 HV for St37 and 200-235 HV for St52, which are higher than those of the dissimilar joints.

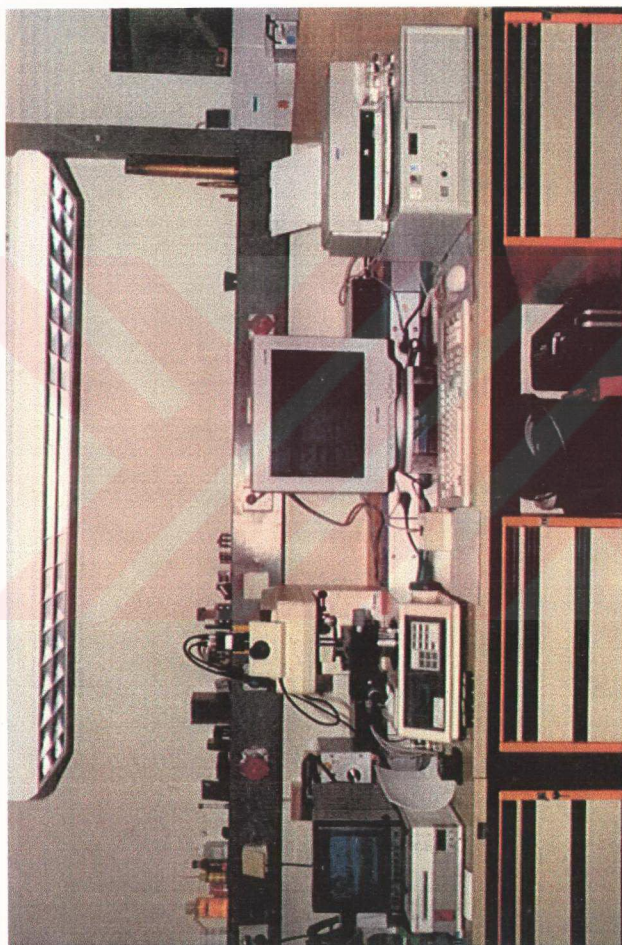


Fig. 4.8 Set-up used for measuring the hardness profiles

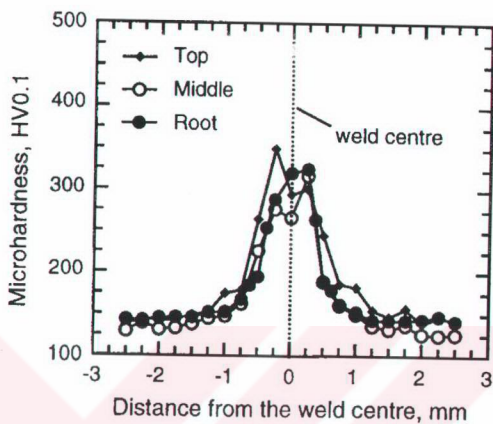


Fig. 4.9 Hardness profile of the St37 similar joint

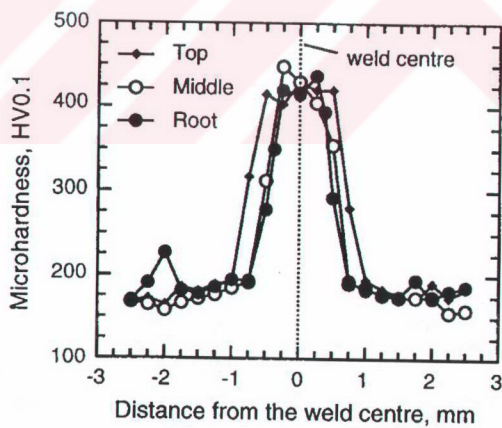


Fig. 4.10 Hardness profile of the St52 similar joint

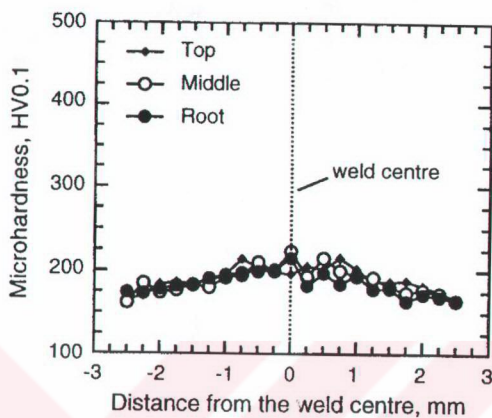


Fig. 4.11 Hardness profile of the aust similar joint

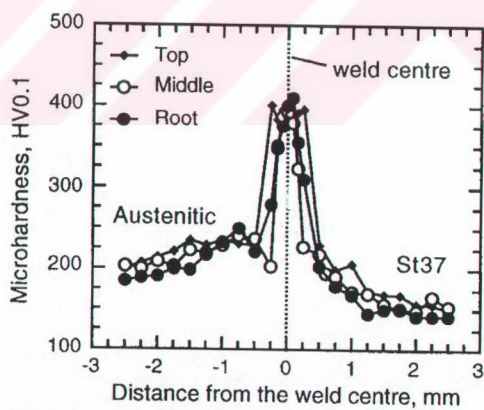


Fig. 4.12 Hardness profile of the St37-aust dissimilar joint

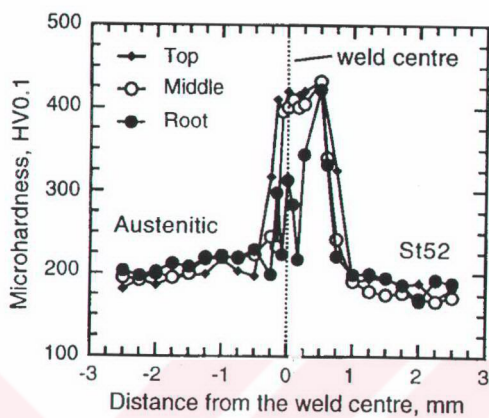


Fig. 4.13 Hardness profile of the St52-aust dissimilar joint

4.3 Mechanical Properties

Evaluation of the mechanical properties is crucial but also very difficult. Firstly, measurement of the actual properties is difficult since the established techniques cannot always be applied without modification. Secondly, the acceptance criteria may consequently have to be modified, which makes it difficult directly to build on the decades of experience connected to this criteria. For example, in narrow laser welds determination of mechanical properties is difficult due to the high strength mis-match between the base metal and the weld region.

Standard round tensile specimens of the base metals were tested to find out the mechanical properties and mis-match ratios (M) as given in Table 4.1. The stress-strain curves of the round tensile specimens (4 mm diameter, 20 mm gauge length) extracted from the base metals are given in Fig. 4.14. The yield strength mis-match ratios between the two base plates of St37-austenitic and St52-austenitic dissimilar joints, $M = YS_{St37}/YS_{aus}$ and $M = YS_{aus}/YS_{St52}$, were found to be 0.76 and 0.8, respectively. Standard flat transverse tensile specimens were also tested to find out the mechanical properties of the joints, Fig. 4.15. The similar and dissimilar joints containing ferritic steels always failed in the lower strength (in terms of tensile strength) ferritic base metal due to the very high hardness of the weld region. The stress-strain curves obtained from flat transverse tensile specimens show some differences compared to the round tensile specimens due to the presence of LB weld region at the middle of the specimens, Figs. 4.14 and 4.15. The stress-strain curves of the dissimilar joints lie between those of the similar joints of the constituent steels as expected, Fig. 4.15. The stress-strain curves obtained from standard round and flat microtensile specimens of the base metals are compared in Fig. 4.16. Although there are slight differences in the curves due to complete different geometry of these two specimen types, they are in good agreement.

The problem in tensile testing of whole mis-matched joints in the transverse direction using either round or standard transverse tensile specimens, is verifying the

soundness of the weld deposit because of the difference in strength and ductility existing between the base and the weld metal. This strength mis-match will result in a failure in the lower strength base metal (overmatching case) even though the weld metal might contain defects. Although a fracture in the base metal can be considered acceptable, a true evaluation of weld metals is not possible under such circumstances. A complete picture of the strength level and its variation within the weld volume should be determined by using “all-weld-metal” and “all-HAZ” specimens.

Therefore, all-weld metal and all-HAZ flat microtensile specimens were prepared and tested to determine the local mechanical properties of the respective areas in the weld region. Table 4.2 summarizes the mechanical properties of the ferritic base metals (BM-F), austenitic base metals (BM-A), weld metals (WM), HAZ's of ferritic steels (HAZ-F) and HAZ of austenitic steel (HAZ-A) on similar and dissimilar joints determined using microtensile specimens. The weld metal regions of the joints containing ferritic constituent exhibit high strength (overmatching) and low strain values whereas austenitic similar joint shows similar strength (even-matching) and strain levels in the weld metal, HAZ and base metal regions as microhardness values indicated. The strength mis-match ratios between weld metal region and respective base metals (lower strength base material in the case of dissimilar joints), $M = YS_{WM}/YS_{BM}$, are also given in Table 4.2. The highest mis-match ($M=3.09$) exists between the weld metal and base metal of similar St37-St37 joint. The full stress-strain curves of these specimens are given in Fig. 4.17. The individual values of the yield strength (YS), tensile strength (TS) and fracture strain with respect to the specimen location for all the joints are presented in Fig. 4.18. The sampled microstructure of each microtensile specimen was checked (after testing) by optical metallography at the gauge length region of the specimens. Figs. 4.19-4.23 are presenting the representative micrographs for all five weld joints. These results are discussed below in detail.

4.3.1 Ferritic Similar Joints

Figs. 4.17a and e illustrate the stress-strain curves of the microtensile specimens extracted from the similar ferritic joints and Figs. 4.18a and e show the variation in the mechanical properties of these joints. As seen from these figures, the weld metals, particularly in the case of St52, exhibit the highest strength levels and the lowest strain values compared to the BM and HAZ regions. The HAZ regions of these joints also exhibited higher strength levels and lower strain values than the respective base metals.

The microstructures of the weld metal, HAZ and base metal specimens of the similar St37 joint are shown in Fig. 4.19. The base metal specimens' microstructure consists of coarse ferrite with some pearlite, Fig. 4.19a. The HAZ microtensile specimen displayed a similar microstructure of ferrite and pearlite containing finer regions, Fig. 4.19b, indicating that the specimen was extracted from the base metal side of the HAZ. The weld metal specimen, on the other hand, exhibited a microstructure comprising grain boundary ferrite with bainite and possibly martensite, Fig. 4.19c. The fracture surfaces of the all-HAZ (extracted from the HAZ) and all-WM (extracted from the weld metal) microtensile specimens of similar St37 joint exhibited a predominantly ductile failure, Fig. 4.24. The fracture surface of the all-HAZ specimen displays an extensive dimple formation, Fig. 4.24a, whereas that of the all-WM specimen exhibits the formation of rather large and limited number of voids and their coalescence, Fig. 4.24b.

The microstructure of the weld metal, HAZ and base metal specimens of the similar St52 joint are shown in Fig. 4.20. The base metal specimen microstructure consists of coarse ferrite with elongated pearlite, Fig. 4.20a. The HAZ microtensile specimen displayed a finer microstructure than the base metal with pearlite dissolution, Fig. 4.20b. The weld metal specimen, on the other hand, exhibited a microstructure comprising martensite and bainite with some grain boundary ferrite, Fig. 4.20c. Fracture surfaces of the microtensile specimens extracted from the similar

ferritic joint exhibited a predominantly ductile failure. Fig. 4.25 shows the fracture surfaces of the all-HAZ and all-WM specimens. The fracture surface of the all-HAZ specimen displays a mixed brittle+ductile fracture mode (regions with flat, brittle fracture appearance corresponding to the pearlite dissolution bands shown in Fig. 4.20b), Fig. 4.25a, whereas that of the all-WM specimen exhibits a predominantly brittle fracture mode, Fig. 4.25b.

4.3.2 Austenitic Similar Joint

For austenitic similar joint, the weld metal and HAZ exhibit slightly lower strain values with no significant alteration in the strength levels, Figs. 4.17c and 4.18c. The microstructures of the base metal, HAZ and weld metal specimens of similar austenitic joint are shown in Fig. 4.21. The base metal and HAZ microtensile specimens exhibited similar microstructures consisting of austenite grains with some delta ferrite, pointing out that there is no appreciable alteration in the microstructure in the HAZ of austenitic steel joints, Figs. 4.21a and b, respectively. The weld metal specimen displayed a homogeneous dendritic structure, Fig. 4.21c. Fracture surfaces of the microtensile specimens extracted from the similar austenitic joint also exhibited a predominantly ductile failure. Fig. 4.26 shows the fracture surfaces of the all-HAZ and all-WM specimens. The fracture surface of the all-HAZ specimen displays an extensive dimple formation, Fig. 4.26a, whereas that of the all-WM specimen exhibits the formation of limited and coarse voids and their coalescence, Fig. 4.26b.

4.3.3 Dissimilar Joints

In dissimilar joints, the weld metals exhibit highest strength levels with limited strain values as seen in Figs. 4.17b and 4.18b and Figs. 4.17d and 4.18d. The reason for this extremely low strain values exhibited by the weld metals is the presence of harder microstructural constituents (bainite/martensite) as well as the solidification cracks in these regions as seen in Fig. 4.6. The HAZs at the ferritic sides (HAZ-F)

also display relatively higher strength levels and lower strain values than the respective ferritic base metals. The HAZs at the austenitic sides (HAZ-A), on the other hand, do not exhibit a significant increase in strength, but some decrease in strain value as compared to the austenitic base metal.

The microstructures of the weld metal, HAZ and base metal specimens of dissimilar St37-austenitic joint are illustrated in Fig. 4.22. The ferritic base metal specimens exhibited a microstructure comprising ferrite and some pearlite. However, the microstructure is significantly finer than that of the base metal of ferritic similar joint, which is the reason for the different stress-strain curves obtained, Figs. 4.17a and b. The difference in the strength of St37 base metals of similar and dissimilar joints is also shown in Table 4.2. This difference is due to the different welding directions with respect to the rolling directions in similar and dissimilar joints of St37 grade. The HAZ at the ferritic side (HAZ-F) exhibited a much finer ferrite+pearlite structure than the ferritic base metal. The weld metal specimen, on the other hand, displayed an inhomogeneous structure containing a mixture of ferrite and austenite solidification structures. The HAZ at the austenitic side (HAZ-A) and austenitic base metal displayed a similar microstructure to those microstructures shown in Figs.4.21a and b.

The microstructures of the weld metal, HAZ and base metal specimens of dissimilar St52-austenitic joint are illustrated in Fig. 4.23. The ferritic base metal specimens exhibited a microstructure comprising ferrite with elongated pearlite, Fig. 4.23a. The HAZ at the ferritic side (HAZ-F) exhibited a finer microstructure than the ferritic base metal with pearlite dissolution, Fig. 4.23b. Although a very narrow region of martensite was observed in the HAZ-F of this joints also indicated by hardness traverse, the microtensile specimen microstructure did not reveal this phase while this small region was not sampled in the extraction of the specimens. The weld metal specimen, on the other hand, displayed an inhomogeneous structure containing a mixture of ferrite and austenite solidification structures. The HAZ at the austenitic

side (HAZ-A) and austenitic base metal displayed a similar microstructure to those microstructures shown in Figs. 4.21a and b.

Fracture surfaces of the microtensile specimens extracted from the base metals and the HAZs of the dissimilar joints also exhibited a predominantly ductile failure. However, the all-WM specimens exhibited a non-uniform fracture surface, which consisted of ductile regions and cleavage fracture of the dendritic zones, Fig. 4.27. Fig. 4.27a shows the fracture surface of the dendritic region where some microplastic zones can be clearly seen. Fig. 4.27b illustrates the corresponding optical micrograph of the dendritic structure formed in the weld zone.

As demonstrated above, the mechanical properties (including full stress-strain curves) of each zone of laser beam welded similar and dissimilar joints can be successfully determined using flat microtensile specimen technique. A complete information (not only hardness values) on the local mechanical properties of laser welds is often essential for optimization of the laser welding process and filler wire development for various alloys as well as quality control and fracture mechanics analyses (experimental and numerical) of the welds. Development of filler wire composition to reduce the level of hardenability of C-Mn steel joints and the level of softening of Al-alloy weld joints are currently of great interest. In order to make a step forward in these areas, a complete information on the local properties of real weld joints should be generated. The testing technique described in this study can provide such an information.

The existing correlation between hardness and strength of C-Mn steel weldments which were originally developed for conventional fusion welds may not be suitable for laser welds. A plot of the hardness vs. yield strength (YS) values of the specimens extracted from the base metal, the HAZ and the weld metal of the five weld combinations are shown as an example in Fig. 4.28a-e. In order to determine the best correlation between hardness and strength values of laser welded steel joints further data is needed. This subject is thought as a future work.

It is seen that transverse tensile test specimens invariably fail in the base material with good overall elongation due to the heavy overmatching. No plastic deformation is occurring in the weld zone, which may give the impression that the ductility of the weld is satisfying. Furthermore, fracture mechanics calculations and tests have shown that a weld metal overmatching in strength protects the weld metal itself as well as its surroundings from developing cracks during transverse stress situations. This way of deformation will, however, not always be achieved in service, as longitudinal deformation of a weld will occur in some situations. It is therefore, relevant to test the weld in the longitudinal direction also. For longitudinal loading conditions, plastic deformation will be controlled by the dominating amount of base material, and the deformation will therefore, become uniformly distributed along the weld. A longitudinal bend test is therefore well-suited for measuring this ductility. In contrast, a longitudinal tensile test will normally show more modest figures for the ductility, as the maximum loading point is met at a rather low strain beyond which the deformation is non-uniform. Alternatively a medium scale longitudinal tensile test containing a larger amount of base metal may be used to assess the weld metal ductility up to the limit of uniform deformation. For the bending test as well as for the medium scale test conditions due to weld metal imperfections (for example pores) must be distinguished from ductile cracks.

It is thought that carrying out longitudinal tensile and bending tests of laser welded specimens would be a suitable research subject in a future work.

4.3.4 Photoelastic Transverse Tensile Specimens

A number of standard transverse tensile specimens from each weld group were photoelastically coated in order to observe the deformation patterns developing during loading. Fig. 4.29 shows the deformation patterns of all the weld joints studied. In the case of St37 similar joint, Fig. 4.29a, the applied stress causes the spread of yielding in the direction of about 45° to the tensile axis on both sides of the weld on the base metal side. As the value of stress is further increased, strains are

distributed at the base metal and the middle part of the weld zone. Final fracture occurs at the St37 base metal. This is a natural consequence since there is a great mis-match ($M=Y_{S_{WM}}/Y_{S_{BM}}=3.09$) between the St37 base metal and the weld region. For St52 similar joint, Fig. 4.29b, although the initial strains develop at the weld region, as the load is increased the strains are distributed in the base metal and final fracture occurs at the lower strength ($M=2.5$) ferritic base metal. In austenitic similar joint, Fig. 4.29c, the deformation begins at the entire base metal surface and since the yield strengths of the weld metal and base metal are close ($M=1.13$), as the applied load increases the deformation covers the whole specimen surface. Fracture occurs at the weld metal since its tensile stress is lower than that of the base metal.

In St37-austenitic dissimilar joint, Fig. 4.29d, deformation begins at the ferritic base metal. At this stage, the weld region and austenitic base metal are nondeformable. As the applied load is increased, the deformation is distributed all over the ferritic and the austenitic base metal and final fracture occurs in the ferritic base metal side. In the case of St52-austenitic dissimilar joint, Fig. 4.29e, the initial yielding occurs at the austenitic base metal side since its yield strength (318 MPa) is lower than that of the St52 (382 MPa). The weld region is initially nondeformable. As the load increases plastic straining shifts through the ferritic base metal and final fracture occurs in this region, since the tensile strength of St52 (528 MPa) is lower than that of the austenitic steel (626 MPa).

Table 4.1 Mechanical properties of the base metals

Base Material	Yield Strength (MPa)	Tensile Strength (MPa)	Elongation %	Reduction in area %
St37	233	368	36	51.06
St52	382	528	30	50.58
Austenite (1.4404)	308	626	69	49.61

Mis-match Ratios Based on the Strength Levels of Base Plates

$$M=Y_{S_{St37}}/Y_{S_{aust}}=233/308=0.76$$

Table 4.2. Summary of BM, HAZ and WM mechanical properties of the laser beam welded joints determined by microtensile specimens. In parenthesis, average values are presented for base metals and weld metals.

	St37-St37			St37-Austenitic			Austenitic-Austenitic			St52-Austenitic			St52-St52		
	YS (MPa)	TS (MPa)		YS (MPa)	TS (MPa)		YS (MPa)	TS (MPa)		YS (MPa)	TS (MPa)		YS (MPa)	TS (MPa)	
BM-F	230	370		330	419					355	545		369	523	
	215 (222)	376 (372)		327 (325)	404 (415)		--	--		372 (363)	522 (535)		379 (372)	536 (536)	
	215	367		334	440					371	529		384	541	
	228	374		307	398					355	545		357	545	
BM-A	--	--		326	593		316	611		312	607		--	--	
				304 (319)	573 (598)		326 (319)	609 (607)		318 (317)	601 (607)				
				320	608		318	601		318	617				
				325	616		316	607		320	602				
WM	640	825					360			745	998		980	1232	
	730 (685)	893 (851)		900	988			552		658 (702)	722 (860)		900 (930)	1141 (1155)	
	638	804		580 (740)	606 (797)								910	1092	
	732	880													
HAZ-F	320	432		416	496								620	870	
	316 (304)	435 (424)		378 (504)	472 (597)		--	--		580	757		564 (484)	760 (682)	
	298	421		718	823					424 (502)	624 (691)		382	557	
	280	406											370	540	
HAZ-A	--	--		424	591		448	621		438	668		--	--	
				410 (417)	613 (602)		430 (439)	619 (620)		396 (417)	590 (629)				
M	3.09			2.28			1.13			2.21			2.53		

$$M = YS_{WM} / YS_{BM}$$

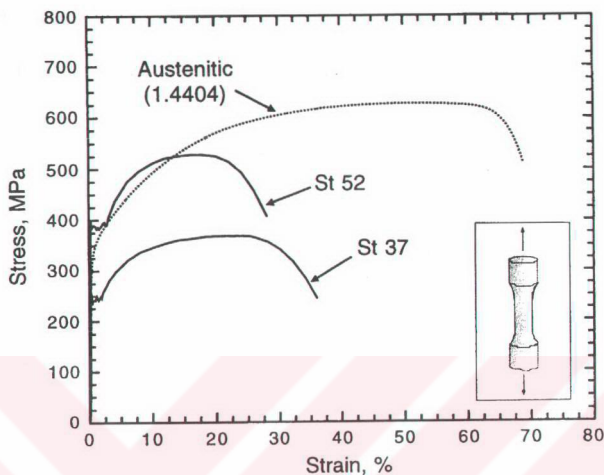


Fig. 4.14 σ - ϵ curves of base metals determined by testing of standard round tensile specimens

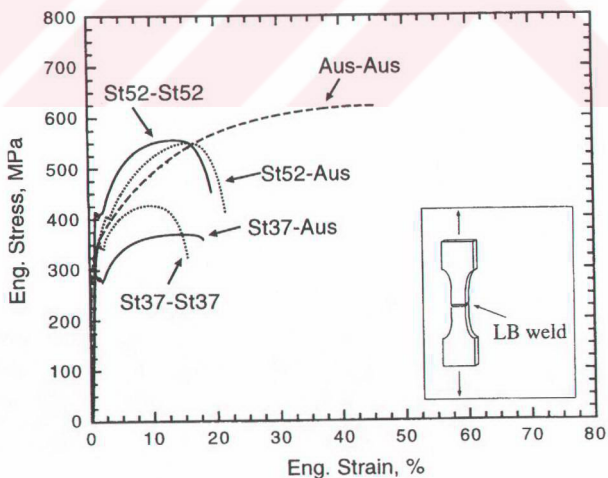


Fig. 4.15 σ - ϵ curves of LB weld joints determined by testing of transverse tensile specimens

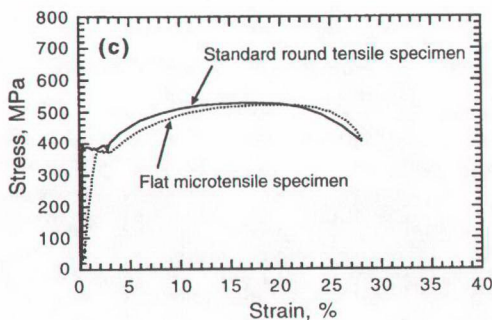
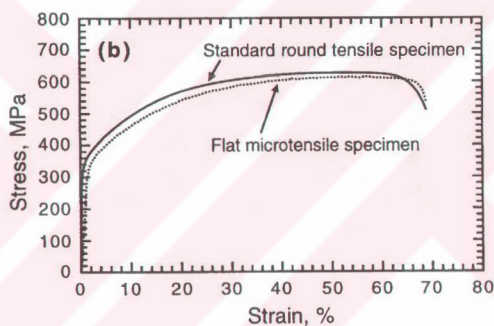
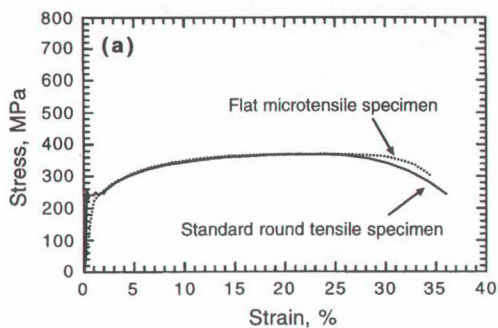


Fig. 4.16 Comparison of the σ - ϵ curves of base metals obtained by testing standard round and flat microtensile specimens **a)** St37, **b)** St52, **c)** austenitic steels

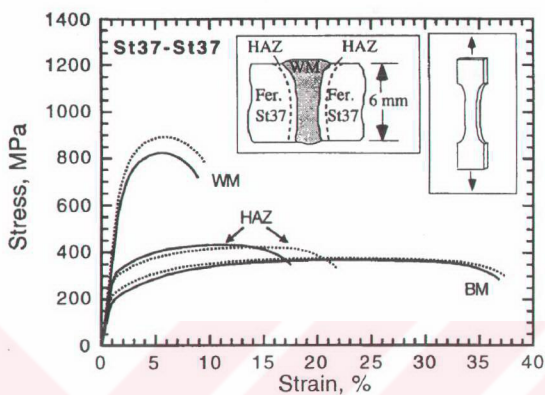


Fig. 4.17 a)

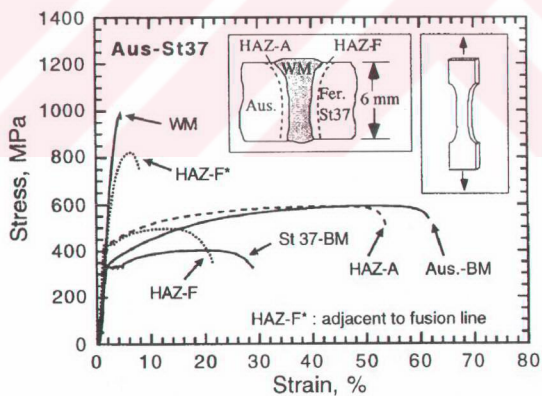


Fig. 4.17 b) (to be continued)

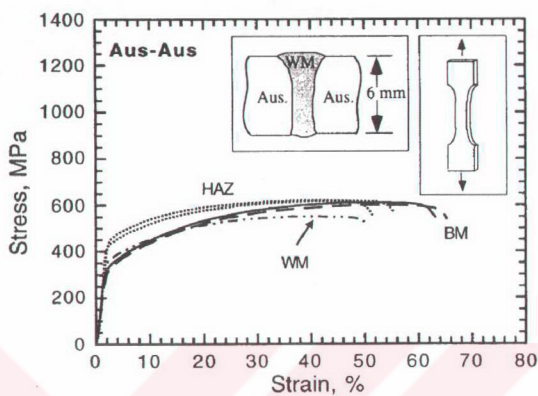


Fig. 4.17 c)

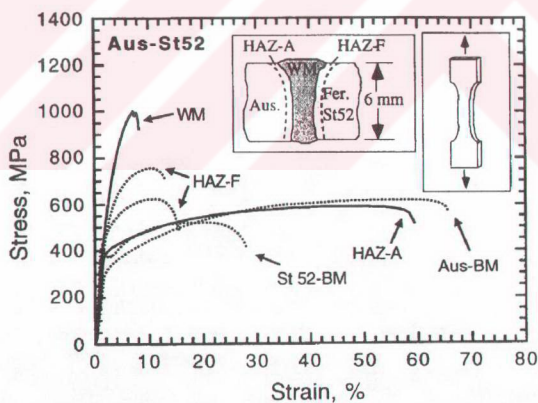


Fig. 4.17 d) (to be continued)

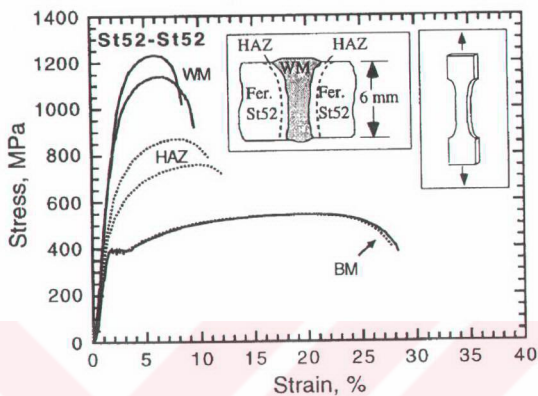


Fig. 4.17 e)

Fig. 4.17 σ - ϵ curves of the flat microtensile specimens extracted from: a) St37 similar, b) St37-aust dissimilar, c) aust similar, d) St52-aust dissimilar, e) St52 similar joints

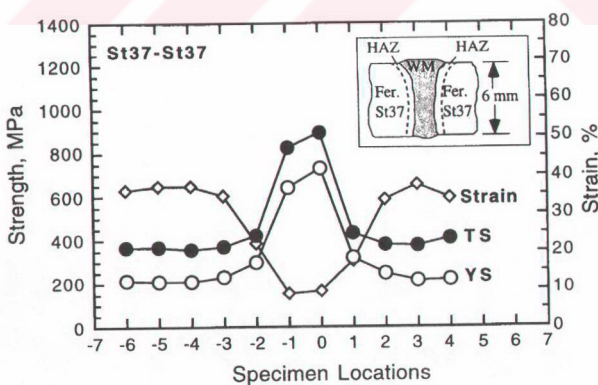


Fig. 4.18 a) (to be continued)

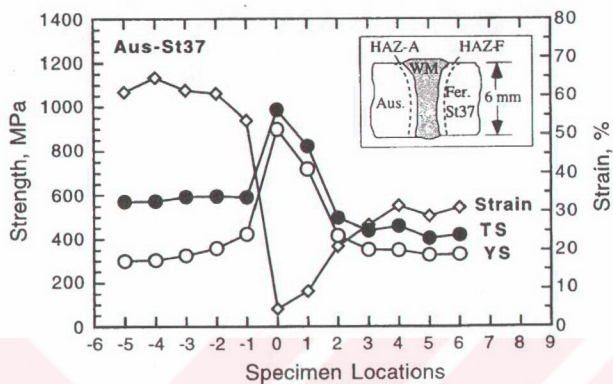


Fig. 4.18 b)

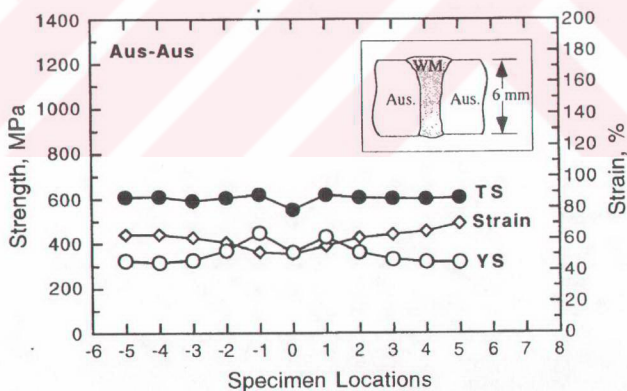


Fig. 4.18 c) (to be continued)

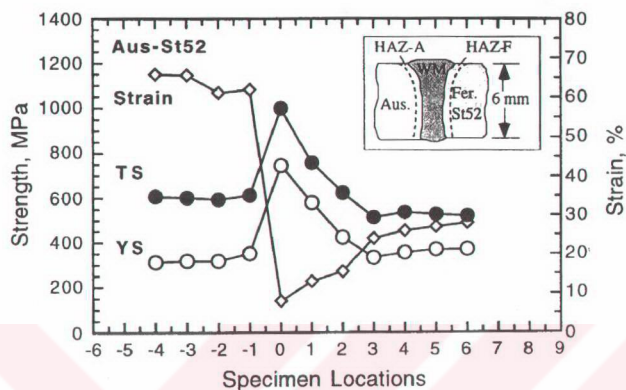


Fig. 4.18 d)

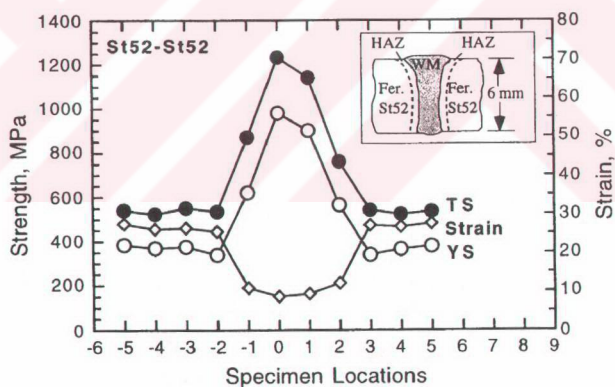


Fig. 4.18 e)

Fig. 4.18 Mechanical property variations across the joints: a) St37 similar, b) St37-aust dissimilar, c) aust similar, d) St52-aust dissimilar, e) St52 similar joints

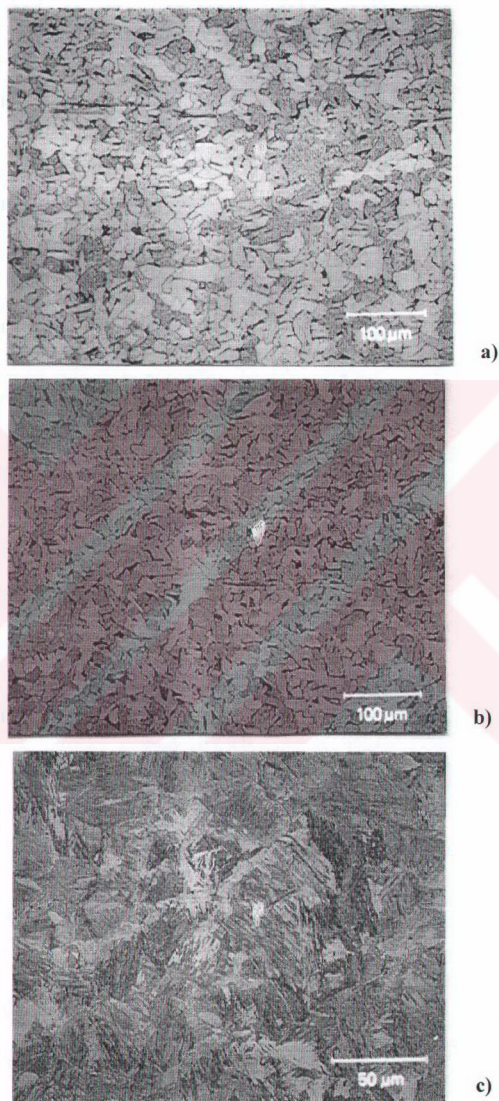


Fig. 4.19 Optical micrographs of the microtensile specimens of St37 similar joint extracted from a) base metal, b) HAZ, c) weld metal

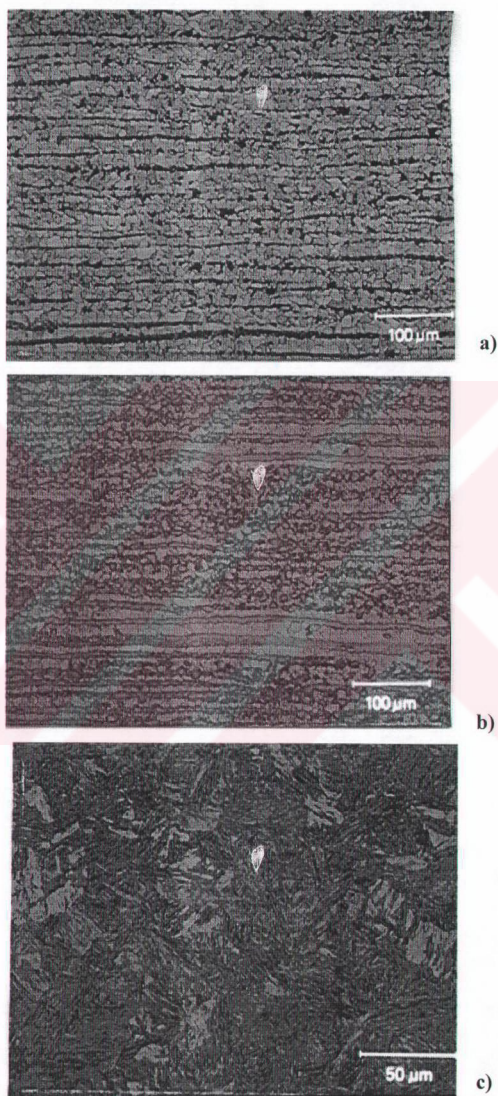


Fig. 4.20 Optical micrographs of the microtensile specimens of St52 similar joint extracted from **a)** base metal, **b)** HAZ, **c)** weld metal

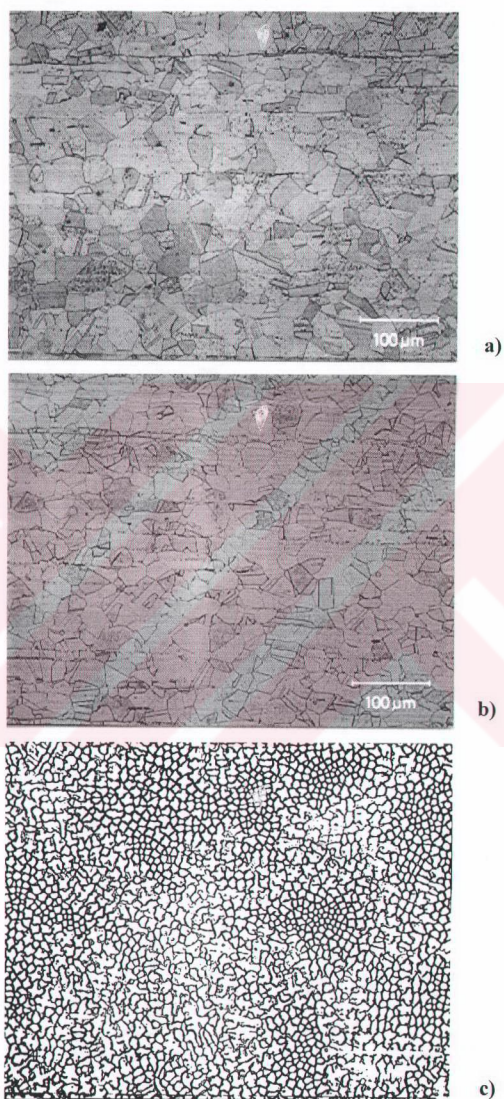


Fig. 4.21 Optical micrographs of the microtensile specimens of aust similar joint extracted from a) base metal, b) HAZ, c) weld metal

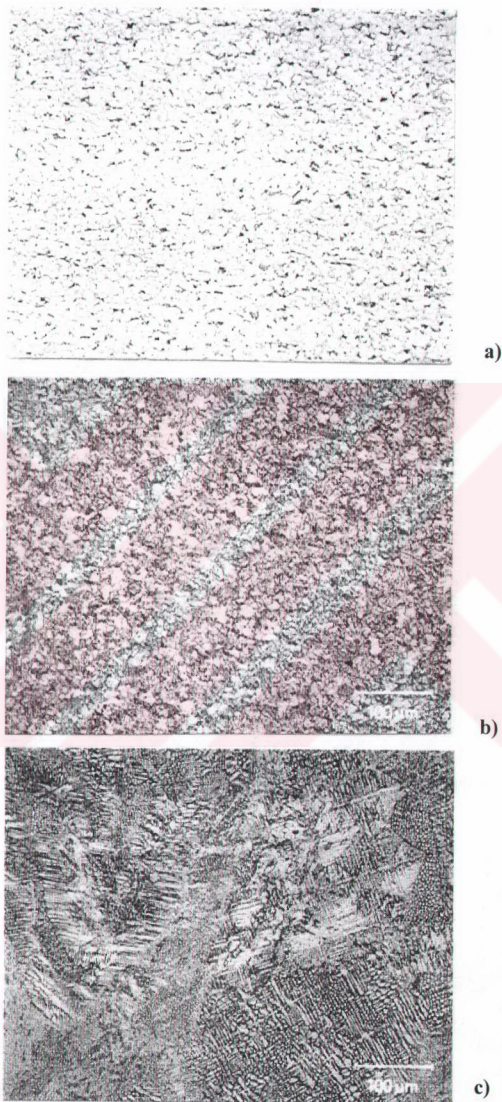


Fig. 4.22 Optical micrographs of the microtensile specimens of St37-aust dissimilar joint extracted from a) St37 base metal, b) St37 HAZ, c) weld metal

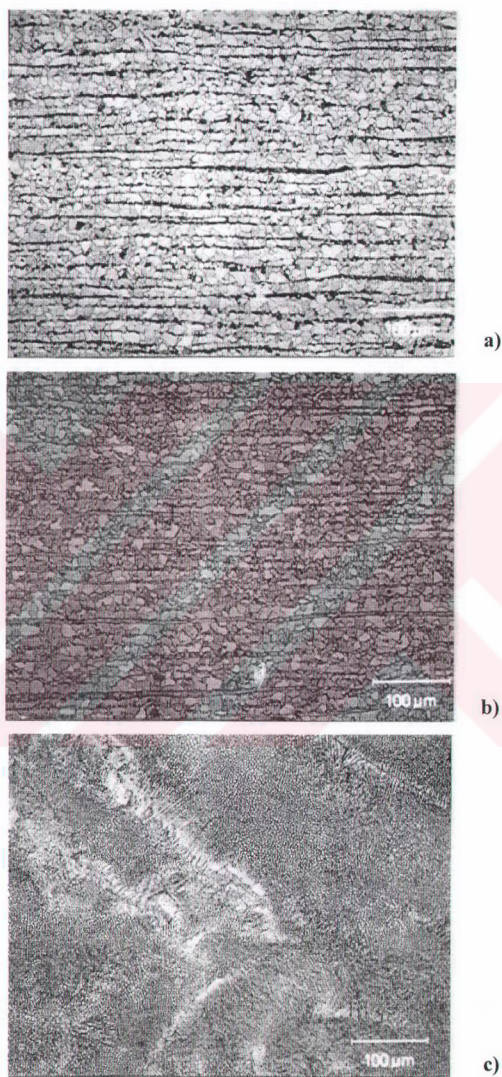
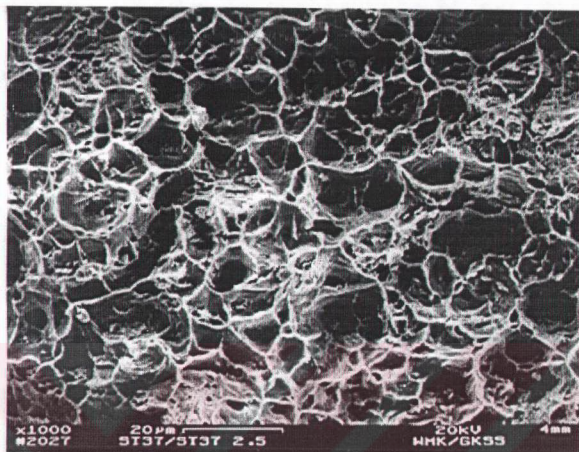
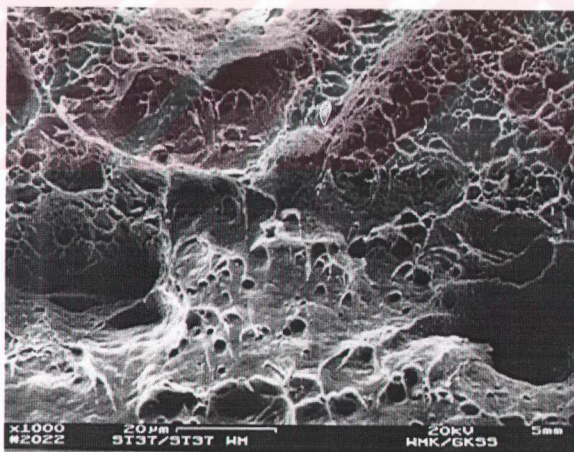


Fig. 4.23 Optical micrographs of the microtensile specimens of St52-aust dissimilar joint extracted from **a)** base metal, **b)** HAZ, **c)** weld metal



a)



b)

Fig. 4.24 Fracture surfaces of the HAZ and WM of St37 similar joint a) fracture surface of the HAZ showing extensive void formation, b) fracture surface of the WM illustrating limited void formation

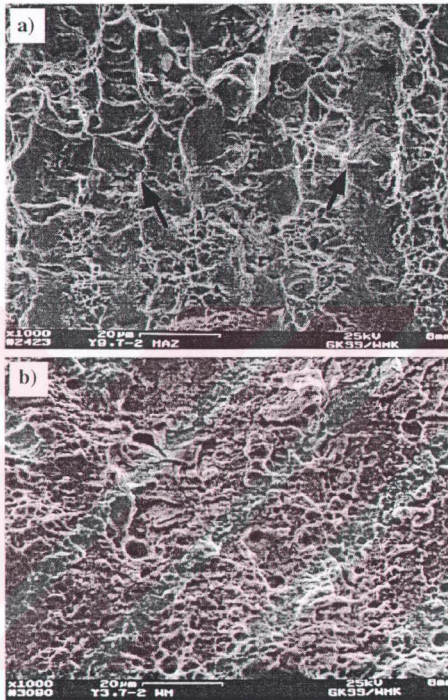
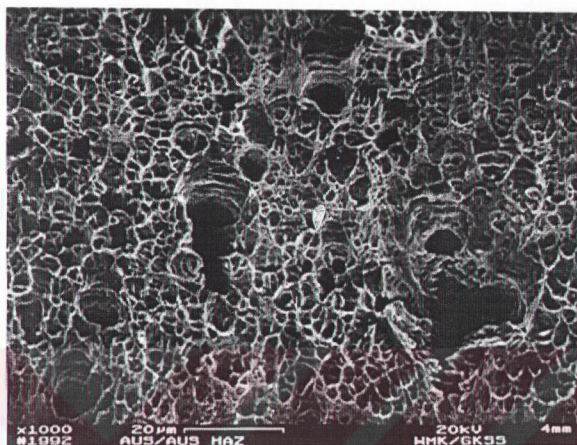
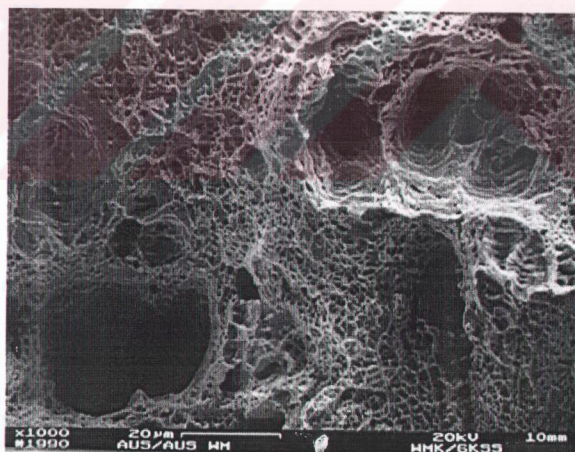


Fig. 4.25 Fracture surfaces of the HAZ and WM of St52 similar joint **a)** fracture surface of the HAZ showing a mixed brittle+ductile mode of fracture (regions with flat, brittle fracture appearance indicated by arrows corresponding to pearlite dissolution bands shown in Fig. 4.20b), **b)** fracture surface of the WM illustrating predominantly brittle fracture mode



a)



b)

Fig. 4.26 Fracture surfaces of the HAZ and WM of aust similar joint a) fracture surface of the HAZ showing extensive void formation, b) fracture surface of the WM illustrating limited void formation

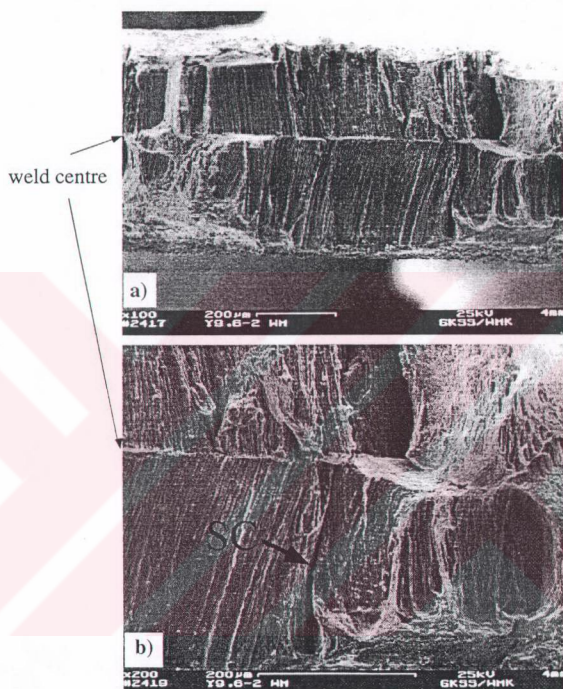
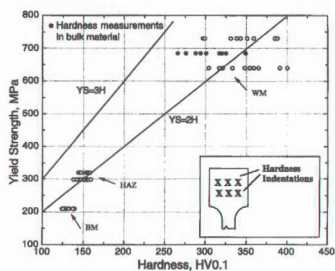
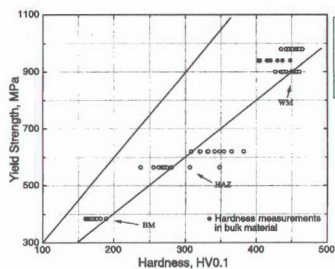


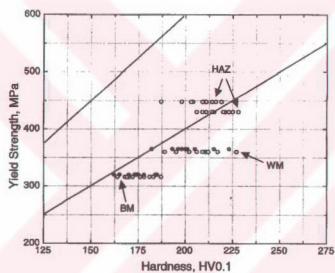
Fig. 4.27 Fracture surfaces of the WM of the dissimilar joint **a)** overview showing the complete thickness of the microtensile specimen, **b)** higher magnification of **a)** illustrating the existing solidification cracking (SC) in the weld metal



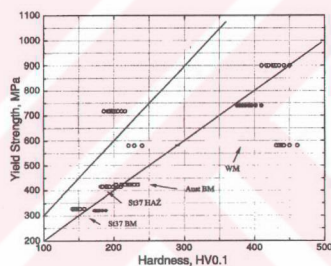
a)



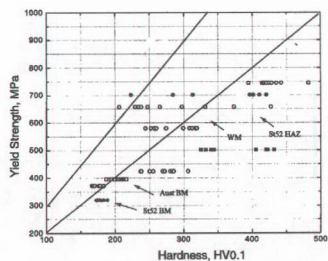
b)



c)



d)



e)

Fig. 4.28 Hardness vs. Yield strength graphics of a) St37 similar, b) St52 similar, c) austenitic similar, d) St37-aust dissimilar, e) St52-aust dissimilar joints

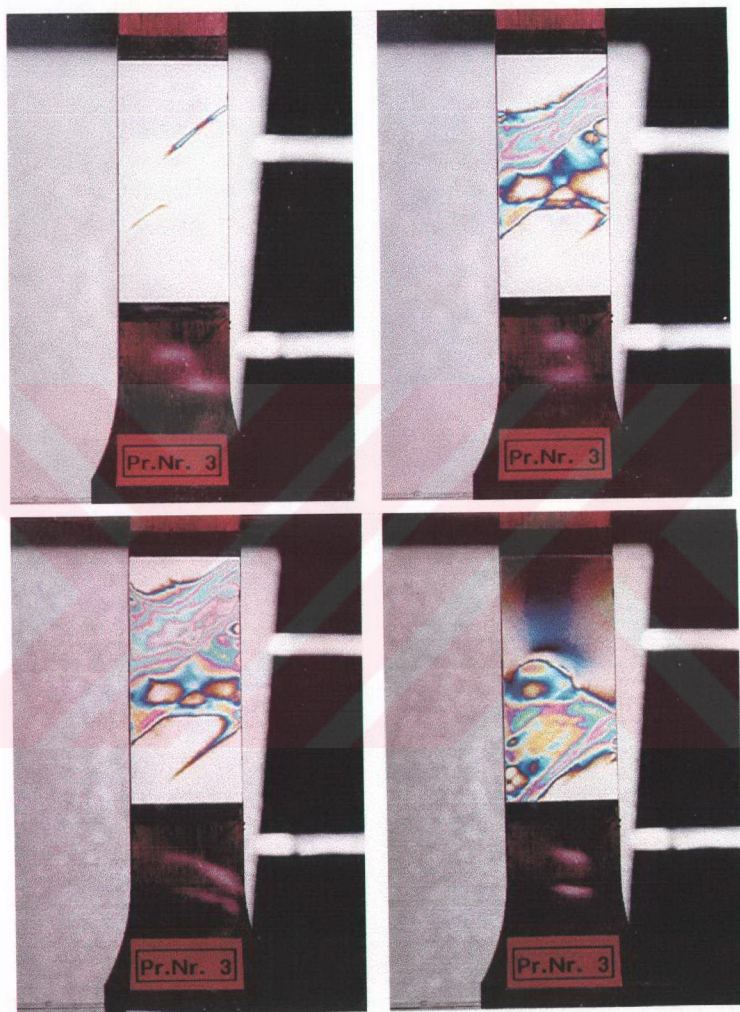


Fig. 4.29 a) (to be continued)



Fig. 4.29 b) (to be continued)



Fig. 4.29 c) (to be continued)

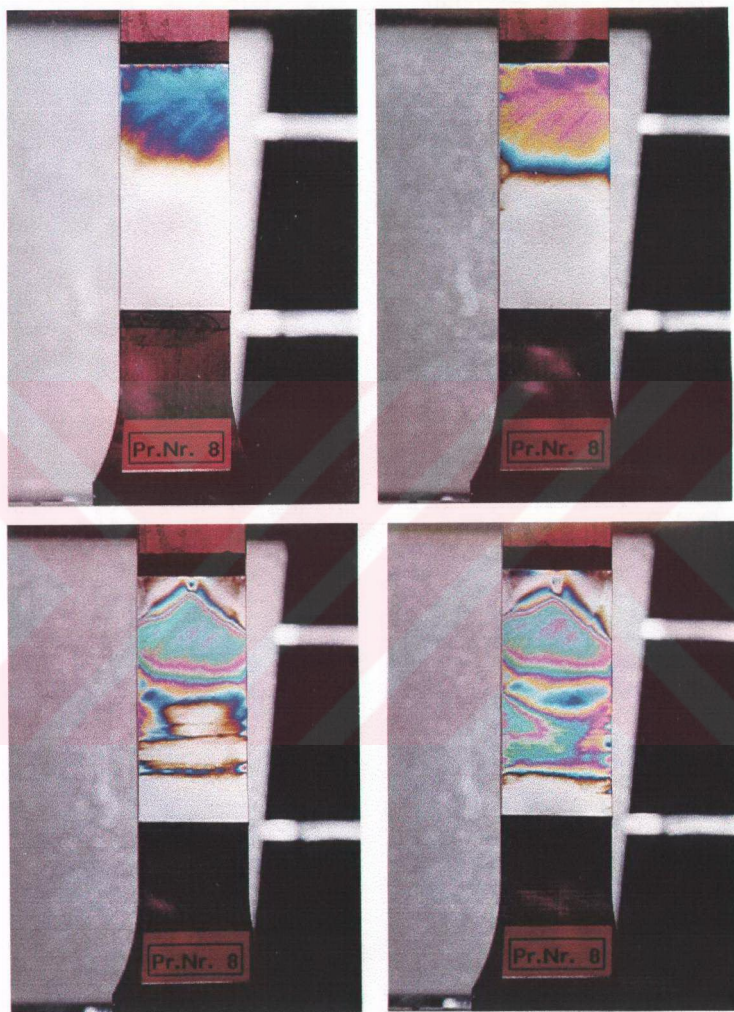


Fig. 4.29 d) (to be continued)

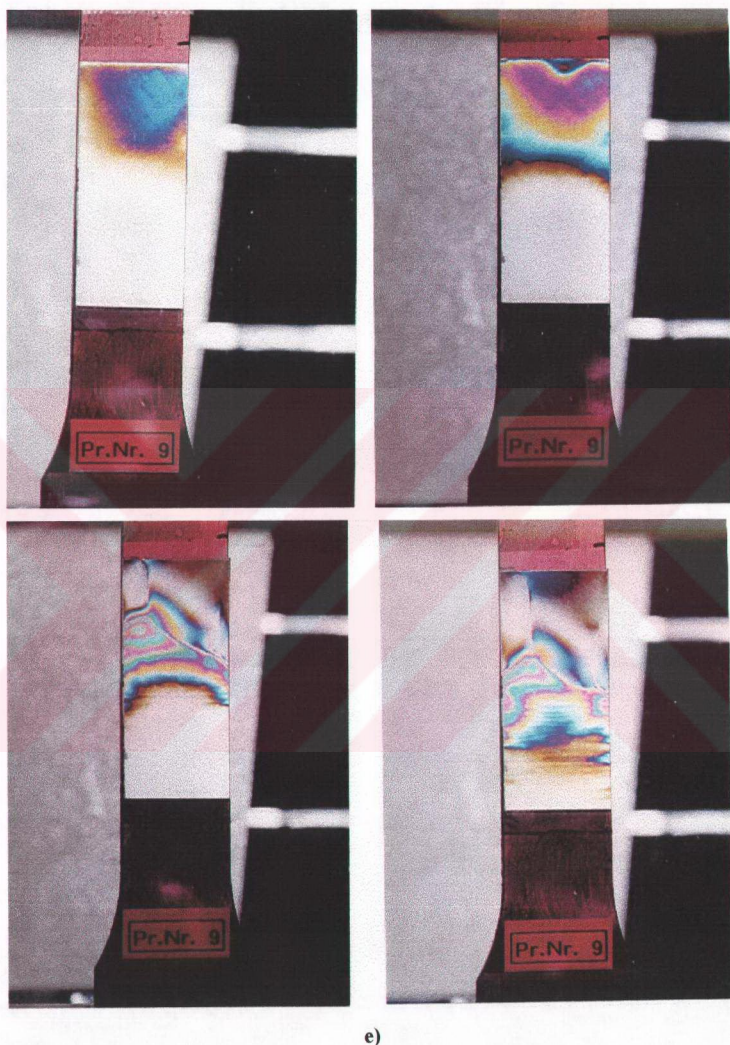


Fig. 4.29 Photoelastically coated transverse tensile specimens showing deformation patterns during loading: **a)** St37 similar, **b)** St52 similar, **c)** austenitic similar, **d)** St37-aust dissimilar, **e)** St52-aust dissimilar joints

4.4 Fracture Toughness Properties

Since the problem of measuring a meaningful fracture toughness value (in terms of standardized CTOD or J) of mis-matched laser weld specimens is very hard to overcome due to the high strength mis-match within very small region at the vicinity of the crack tip, it is not a simple task to distinguish the contributions from both parent materials (lower strength) and narrow LB weld (highly overmatched) at the vicinity of the crack tip to the remotely measured crack mouth opening displacement (CMOD) and load line displacement (V_{LL}) usually used in standardized CTOD and J estimates. Therefore, the local and direct measurement technique (δ_5 technique), developed at GKSS Research Center as a measure of the CTOD for determining the fracture toughness and the crack growth resistance, consists of measuring the relative displacement of two relative gauge points directly at the crack tip using special displacement gauges. The resulting CTOD is called δ_5 because the gauge length over which the CTOD is determined amounts to 5 mm. The CTOD values of the laser weld joints on the SENB specimens have, therefore, been measured in terms of the CTOD (δ_5) technique.

SENB specimens which were machine notched in the fusion zone are further fatigue pre-cracked in order to introduce a sharp crack ($a/W=0.5$). Fracture toughness measurements (CTOD and CMOD) on 6 mm thick and 25 mm wide SENB specimens loaded under three point bending were carried out at room temperature (RT) and $-40\text{ }^{\circ}\text{C}$. Surface notch position and SENB specimen configurations of the joints are given in Fig. 4.30. Photoelastic specimens were prepared to examine the deformation patterns during loading at room temperature tests. CTOD and CMOD values recorded with respect to the applied load are converted into graphics. After the bend testing, SENB specimens were broken by further fatigue cracking. In order to examine the fracture surfaces, broken specimens were cut near the fracture surfaces.

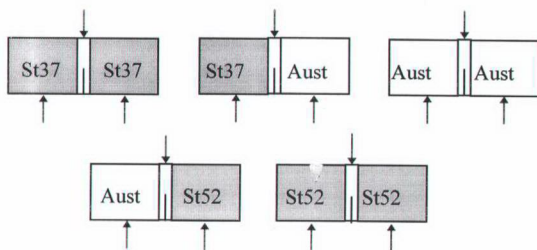


Fig. 4.30 Surface notch position and SENB specimen configurations of the dissimilar joints

For each weld condition, deeply notched ($a/W=0.5$) three point bend specimens were tested. CTOD and CMOD values of three point SENB specimens tested at RT and $-40\text{ }^{\circ}\text{C}$ are given in Table 4.3. The CTOD values given represent the values measured at the maximum load (denoted as δ_m) or the point of pop-in (denoted as δ_p). The CTOD values obtained from these tests are plotted in Fig. 4.31 for RT and Fig. 4.32 for $-40\text{ }^{\circ}\text{C}$.

4.4.1 Room Temperature Tests

All SENB specimens exhibited fully ductile fracture behavior at room temperature tests. The average of the CTOD values for base metal St37 is 1.07, Fig. 4.31a, while for austenitic base metal this value is 2.09, higher as expected since austenitic steel is more ductile than St37 as seen from the stress-strain curves. The average CTOD value for similar St37 is 1.98 and for austenitic similar joint it is 1.33. The increase in CTOD value between St37 BM and similar joint may be attributed to extensive crack deviation from the weld zone into the softer BM (strength overmatching) observed in these joints. The very high CTOD values for similar ferritic weld joint are obviously not representing the intrinsic toughness properties of the weld zone which showed very high hardness values and predominantly martensitic microstructure. In austenitic similar joint, there is significant decrease in CTOD values compared to those of austenitic BM. This may be due to the formation of an austenitic-dendritic (cellular)

inhomogeneous microstructure which contained ferritic-austenitic and bainitic and/or martensitic phases in varying degrees in terms of the similar ferritic joint.

No pop-in behavior was observed in room temperature tests.

4.4.2 -40 °C Tests

Three point bend tests at -40 °C were conducted in order to determine the low temperature behavior of the laser welded joints, hence see the effect of temperature on the fracture behavior of the SENB specimens. Pop-in behavior was observed in -40 °C tests. The pop-in step is usually characterized by an abrupt force drop and a small increase in displacement. It is associated with the arrest of a rapidly moving crack during crack growth.

As seen from Fig. 4.32a, the average CTOD value of the St37 BM is 1.29, which is similar to the CTOD value at RT. The CTOD values correspond to the point of pop-in. An example is given in Fig. 4.33. There is not a significant change in CTOD values due to low temperature. The CTOD value for austenitic BM, measured at the point of pop-in, is 1.49 which is lower than that at RT, indicating the decrease in CTOD with decreasing test temperature. It is seen that the decrease in temperature has apparently resulted in a decrease in toughness of austenitic BM. The average CTOD values of the similar St37 and austenitic joints are 1.25 and 0.86, respectively. The CTOD values correspond to the point of pop-in or the maximum load, whichever is the greatest. The similar CTOD values obtained from similar St37 and St37 BM specimens both at RT and -40 °C can be explained in terms of microstructure developed in the weld zone of St37 laser welded joints which consists of bainite plus grain boundary ferrite apparently having relatively high toughness within this temperature range. The significant decrease at -40 °C in the similar austenitic joint is due to the formation of an austenitic-dendritic (cellular) structure in the weld zone, which may cause a decrease in toughness. The average CTOD result for the St37-austenitic dissimilar joint is 1.08 which indicates that no reduction in CTOD values

at $-40\text{ }^{\circ}\text{C}$ in comparison to RT was observed. The low CTOD value may be attributed to increased constraint and the microstructural effect in the dissimilar joint.

For St52 BM, the average CTOD value is 0.78, Fig. 4.32b, which is similar to the CTOD value at RT. The effect of temperature is not apparent for BM St52. Pop-in was observed in only one of the specimens, which took place at a much lower load than the maximum load. However, as mentioned above the CTOD values of austenitic BM measured at $-40\text{ }^{\circ}\text{C}$ are lower than those at RT, indicating the decrease in CTOD with decreasing test temperature. The average CTOD values for similar St52 and similar austenitic joints at $-40\text{ }^{\circ}\text{C}$ are 0.31 and 0.86, respectively. It is seen that there is a significant decrease in CTOD values for St52 $-40\text{ }^{\circ}\text{C}$ results compared to RT tests. An extreme pop-in behavior observed in these joints; an example is given in Fig. 4.34. A significant reduction in toughness with decreasing temperature is observed which is due to the low toughness of predominantly martensitic microstructure formed in the weld zone. The decrease in the CTOD value of similar austenitic joint with temperature is due to the formation of an austenitic-dendritic (cellular) structure in the weld zone, as described above. The average CTOD value for St52-austenitic dissimilar joint is 0.46. In one specimen, no pop-in was observed, while in the other two specimens several pop-in steps were recorded, Fig. 4.35. Dissimilar St52-austenitic joint exhibited lower CTOD values than similar austenitic specimens at $-40\text{ }^{\circ}\text{C}$. This may be attributed to the increase of constraint (mis-match) due to the introduction of a higher strength material (St52) into the system. On the other hand, the dissimilar joint displayed higher CTOD values than similar St52 joint specimens which is due to the decrease of constraint.

The effect of temperature on CTOD values is given in Fig. 4.36. It is seen that low temperature has a deteriorating effect especially on similar St52 joints Fig. 4.36 b, and to a lesser degree on BM austenitic, similar austenitic, and similar St37 joints, Fig. 4.36a. No significant decrease in CTOD is observed with decreasing temperature on dissimilar joints. Pop-in behavior is observed in almost all of the $-40\text{ }^{\circ}\text{C}$ results, though it is most pronounced in similar St52 specimens.

Fracture surface examinations of the SENB specimens after testing exhibited that there is no distinct crack path deviation into the base metal in the austenitic similar joints both at RT and -40 °C. The crack propagates along the weld region parallel to the weld, Fig. 4.37a, illustrating that these specimens do not contain strength mismatch between laser weld zone and austenitic base metal as hardness results indicated. In the similar ferritic joints, Fig. 4.37 b and c, at both temperatures, the cracks always deviated into the softer base metal due to extensive plastic zone development within the base metal. This was also the case in the dissimilar St37-austenitic joints where the crack path deviated towards the lower strength ferritic side, Fig. 4.37d. Whereas in the dissimilar St52-austenitic case, crack propagation was towards the austenitic BM side, Fig. 4.37e. These situations were expected from the hardness profiles of the joints. The crack path deviation into the soft base metal during standard three point bend tests with the notch in the highly overmatched narrow laser weld zone leads to an experimental difficulty in determination of intrinsic fracture toughness properties of laser welds. The crack propagations in similar and dissimilar LB joints under three point bending condition are schematically shown in Fig. 4.38.

4.4.3 Solidification Cracking

Solidification cracking in the weld zone is caused by the inability of the liquid to support the strain imposed by solidification shrinkage and thermal stresses. High heat inputs, such as high currents and slow welding speeds are believed to contribute to weld solidification cracking. Another factor contributing to solidification cracking is the weld metal grain size. Because, weld metal grain sizes produced by power beam welding processes are usually fine due to rapid cooling involved in these processes. Low heat input power welding processes may thus reduce weld crack sensitivity.

The microstructures of austenitic stainless steel welds typically contain a variety of complex austenite-ferrite structures. These structures are a result of both the solidification behavior and subsequent solid state transformations which are

controlled by both composition and weld cooling rates. Hence, austenitic stainless steels are more susceptible to solidification cracking. However, solidification cracking at room temperature tests has not been observed which might be due to a very fine austenite cell size characteristic of primary austenite solidification of the process of this steel grade.

It is interesting to note that the presence of small solidification cracks in the weld zone of dissimilar joints, Fig. 4.39, did not show any detrimental effect on the fracture toughness values at room temperature tests. This can be due to the fact that extensive plastic deformation developed at the lower strength side of the specimen (shielding effect of the overmatched weld zone). Also it is observed that solidification cracks are not parallel to the crack propagation direction. They are about 45° to the crack path. This could be the reason why there is no opening in the solidification cracks at room temperature tests.

4.4.4 Photoelastic Specimens

One SENB specimen from each weld group was photoelastically coated in order to obtain the deformation patterns occurring at three point bend specimens tested at room temperature.

In St37 similar joint, initial deformation starts at the tip of the crack, Fig. 4.40a and is distributed to ferritic BM at one side of the weld zone. As the load increases it proceeds to BMs at both sides of the weld and it is seen that crack propagates into the lower strength ferritic BM. A similar trend is observed for St52 similar joint, Fig. 4.40b, and for the austenitic similar joint, Fig. 4.40c. In St52 similar joint the crack propagates into the ferritic BM, whereas in austenitic similar joint the crack propagates along the borderline between the BM and the weld region, since the degree of mis-match between the austenitic BM and weld region is the smallest among the weld combinations.

In St37-dissimilar joint, the deformation starts at the crack tip, similar to the other cases, Fig. 4.40d. As the applied load is increased the strains are concentrated mostly on the ferritic side, though, later on, they seem to be evenly distributed to both sides of the weld region. It is observed that crack propagates towards the lower strength ferritic BM. The case in St52-austenitic dissimilar joint is somewhat opposite compared to St37-austenitic joint. Initial deformation occurs at the crack tip, Fig. 4.40e and it first proceeds to the ferritic BM side. As the load increases the deformation is distributed to the austenitic BM side. When the crack tip area is observed, it is seen that crack propagates in the austenitic BM, parallel to the border between the BM and the weld zone.

4.4.5 CMOD Results

Crack Mouth Opening Displacement (CMOD) values were also measured during fracture mechanics tests. CMOD measurements (of standard fracture toughness testing procedure) as made on a part of the crack faces must bear some relation to crack tip displacement even in the bi-material and mis-matched cases, but it is obvious that the nearer displacement measurements are made to the crack tip itself, the more reliable they are likely to be. The CMOD values are included in order to provide further information on the deformation characteristics of SENB specimens, Fig. 4.41a-d; they are corresponding to the point of instability (pop-in) or maximum load attained for each specimen.

A similarity to CTOD vs. temperature values is observed in CMOD vs. Temperature tests, as can be deduced from Fig. 4.42. As seen in the figure, there is no decrease in BM St37 values, Fig. 4.42a, with the decrease in temperature whereas a small decrease is observed in BM austenitic and similar St37 and similar austenitic joints. Similar to the CTOD results no reduction in -40 °C results in dissimilar St37-austenitic joint was observed compared to RT tests which may be attributed to the microstructural effect in the dissimilar joint.

St52-austenitic joint CMOD vs. temperature results are also similar to CTOD vs. temperature results, Fig. 4.42b. St52 BM results are similar at both temperatures. A significant decrease is observed in St52 similar joint which is also the case in CTOD results. An extreme pop-in behavior is observed in this joint as described earlier. Room temperature and -40 °C CMOD results of the dissimilar St52-austenitic joint are alike.

Table 4.3 CTOD&CMOD Values of Three Point SENB Specimens at RT and -40 °C

Material	Test Temp.	Max. Load (kN)	Last Applied Load (kN)	CTOD (mm)	CMOD (mm)
St52 BM	RT	6.36 5.90 (6.13)	6.24 5.78 (6.01)	0.79 0.83 (0.81)	2.31 2.34 (2.33)
St52 BM	-40 °C	7.29 7.12 6.86 (7.09)	7.20 7.06 6.76 (7.01)	0.75 0.91 0.69 (0.78)	2.34 3.29 2.40 (2.68)
St37 BM	RT	4.91 4.94 (4.93)	4.82 4.79 (4.81)	1.04 1.10 (1.07)	3.21 3.12 (3.17)
St37 BM	-40 °C	6.31 5.90 6.02 (6.08)	6.16 5.61 5.78 (5.85)	0.95 1.55 1.38 (1.29)	3.11 4.91 4.26 (4.09)
Aust BM	RT	6.68 6.30 6.90 (6.27)	6.61 6.20 6.82 (6.54)	2.03 1.95 2.30 (2.09)	5.37 5.96 6.90 (6.08)
Aust BM	-40 °C	7.50 8.19 (7.85)	7.29 8.12 (7.71)	1.33 1.64 (1.49)	4.87 5.02 (4.95)
St52-St52	RT	6.91 8.06 6.57 (7.18)	6.82 8.02 6.11 (6.98)	1.43 1.58 0.77 (1.26)	4.24 4.63 2.55 (3.81)
St52-St52	-40 °C	6.37 7.82 5.43 (6.54)	4.10 7.65 5.24 (5.66)	0.16 0.58 0.20 (0.31)	0.73 1.8 0.81 (1.11)
St37-St37	RT	6.13 6.29 5.59 (6.00)	5.88 6.23 5.52 (5.88)	2.26 1.96 1.72 (1.98)	5.53 5.25 4.66 (5.15)
S37-St37	-40 °C	6.94 6.49 6.88 (6.77)	6.65 6.42 6.86 (6.64)	1.04 1.41 1.30 (1.25)	3.52 4.62 4.47 (4.20)
Aust-Aust	RT	6.05 6.04 6.90 (6.33)	5.95 5.94 6.80 (6.23)	1.72 1.17 1.10 (1.33)	5.12 3.76 3.25 (4.04)
Aust-Aust	-40 °C	7.28 7.06 6.48 (6.94)	6.90 7.00 6.39 (6.76)	1.06 0.65 0.86 (0.86)	3.64 2.53 2.93 (3.03)
Aust-St37	RT	6.27 6.64 6.04 (6.32)	5.99 6.50 5.92 (6.14)	0.36 0.95 0.99 (0.77)	1.37 3.40 3.19 (2.65)
Aust-St37	-40 °C	7.44 7.14 7.69 (7.42)	7.41 6.97 7.63 (7.34)	1.30 0.87 1.07 (1.08)	4.79 3.29 3.02 (3.70)
Aust-St52	RT	6.50 6.26 5.42 (6.06)	6.35 5.83 5.23 (5.80)	0.47 0.39 0.51 (0.46)	1.77 1.27 1.62 (1.55)
Aust-St52	-40 °C	7.52 6.44 6.16 (6.71)	7.27 6.36 6.00 (6.54)	0.64 0.51 0.23 (0.46)	2.04 2.00 0.90 (1.65)

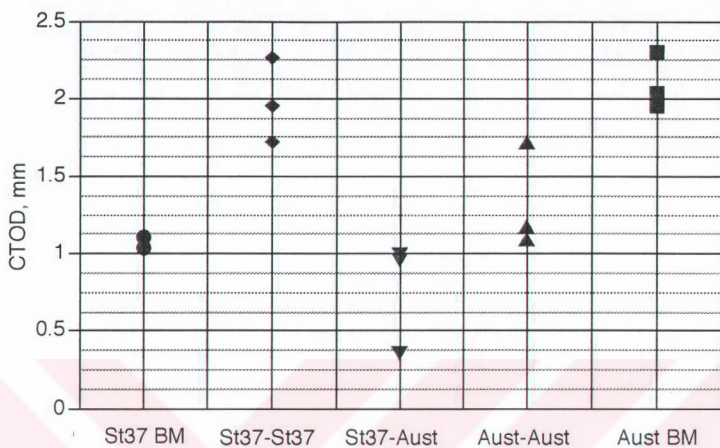


Fig. 4.31 a) CTOD vs. St37 BM, St37 similar, St37-Aust, Aust similar, Aust BM diagrams at RT

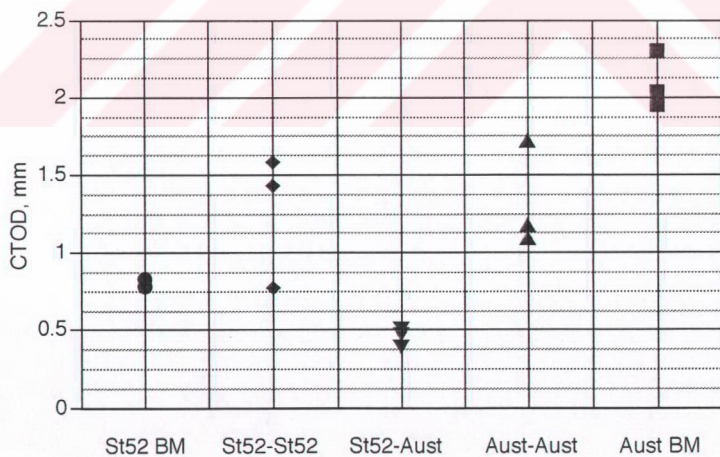


Fig. 4.31 b) CTOD vs. St52 BM, St52 similar, St52-Aust, Aust similar, Aust BM diagrams at RT

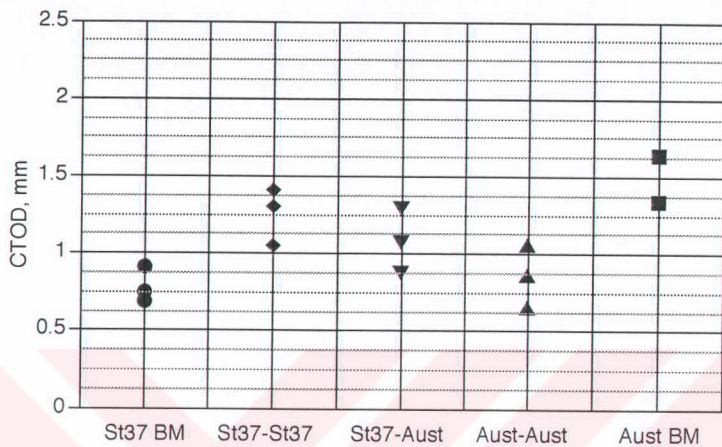


Fig. 4.32 a) CTOD vs. St37 BM, St37 similar, St37-Aust, Aust similar, Aust BM diagrams at -40 °C

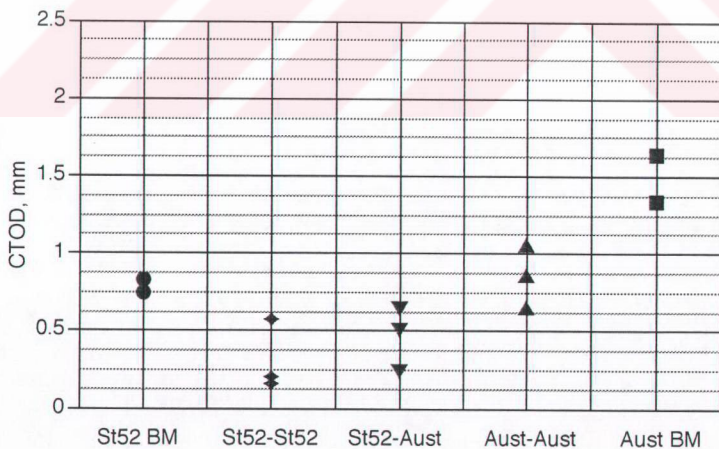


Fig. 4.32 b) CTOD vs. St52 BM, St52 similar, St52-Aust, Aust similar, Aust BM diagrams at -40 °C

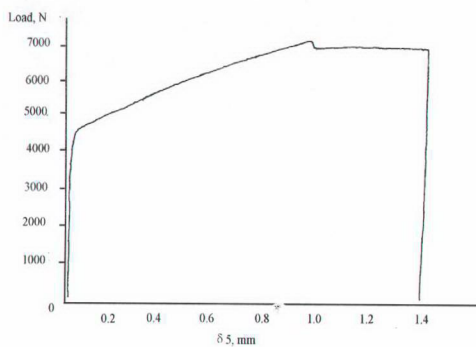


Fig. 4.33 Example of a pop-in behavior at $-40\text{ }^{\circ}\text{C}$

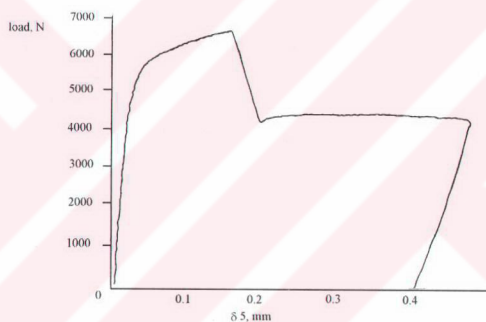


Fig. 4.34 Example of an extreme pop-in behavior at $-40\text{ }^{\circ}\text{C}$

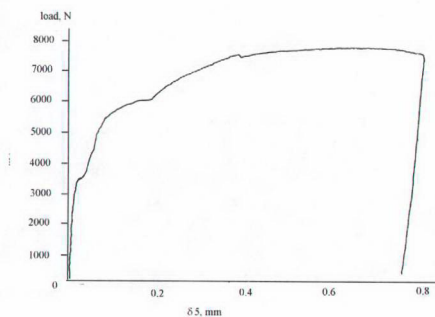


Fig. 4.35 Example of several pop-ins

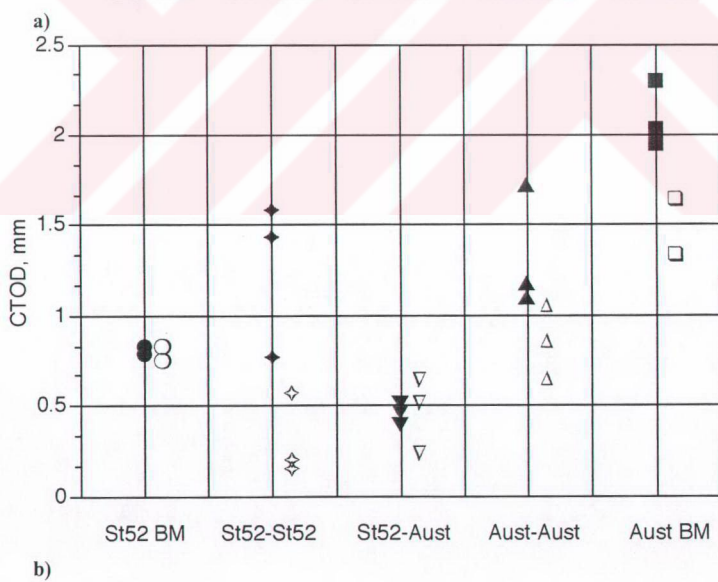
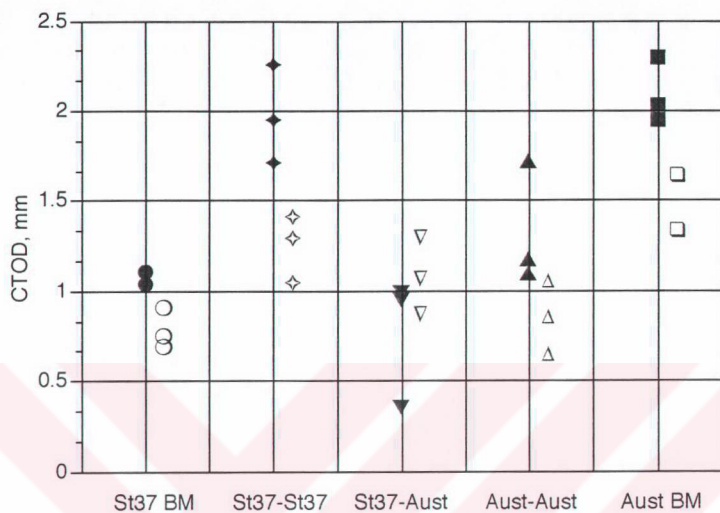


Fig. 4.36 Effect of temperature on CTOD values a) St37-aust joint b) St52-aust joint

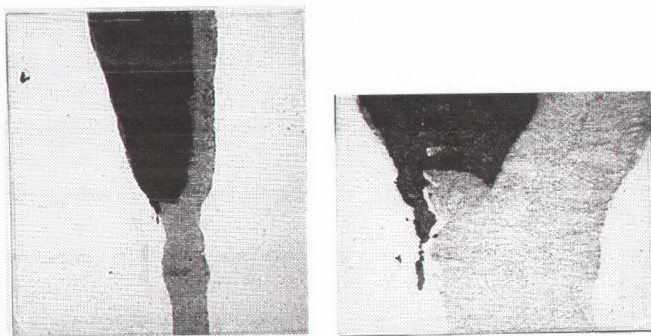


Fig. 4.37 a) Crack path deviation and its higher magnification in Aust similar joint

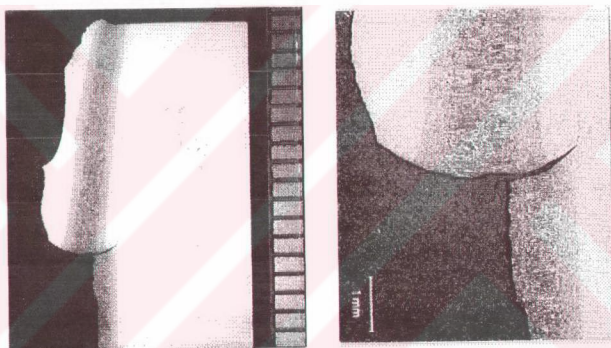


Fig. 4.37 b) Crack path deviation and its higher magnification in St37 similar joint

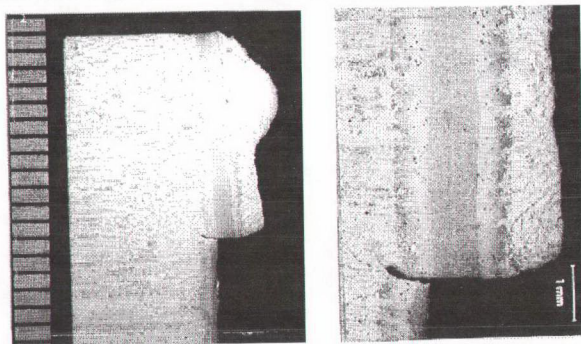


Fig. 4.37 c) Crack path deviation and its higher magnification in St52 similar joint

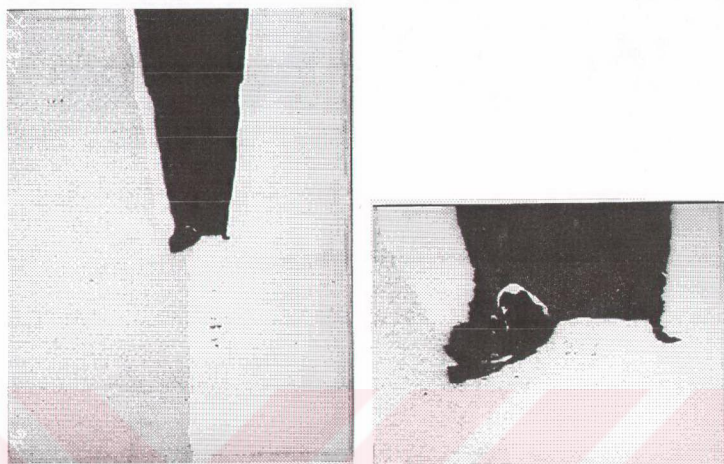


Fig. 4.37 d) Crack path deviation and its higher magnification in St37-Aust dissimilar joint



Fig. 4.37 e) Crack path propagation in St52-Aust dissimilar joint

Fig. 4.37 Crack path propagations & deviations in **a)** Austenitic similar, **b)** St37 similar, **c)** St52 similar, **d)** St37-aust dissimilar, **e)** St52-aust dissimilar joints

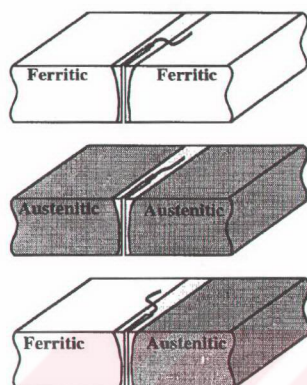


Fig. 4.38 Schematic illustrations of crack propagations in similar and dissimilar joints

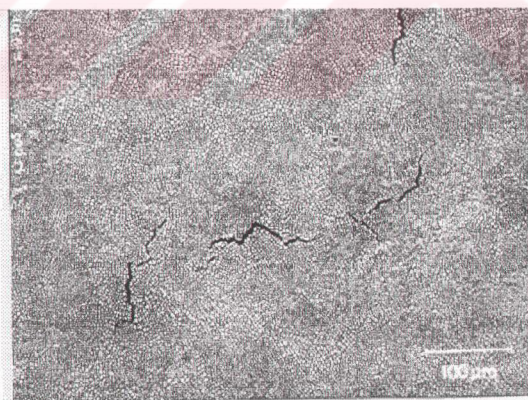


Fig. 4.39 Small solidification cracks in the weld zone of dissimilar joints

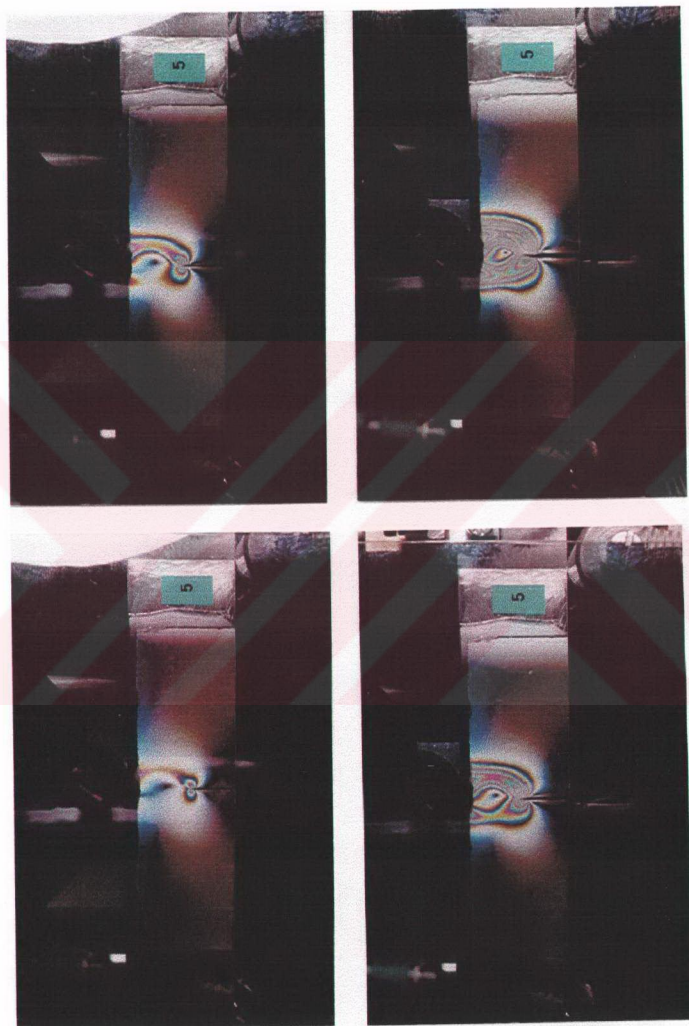


Fig. 4.40 a) St37 similar joint

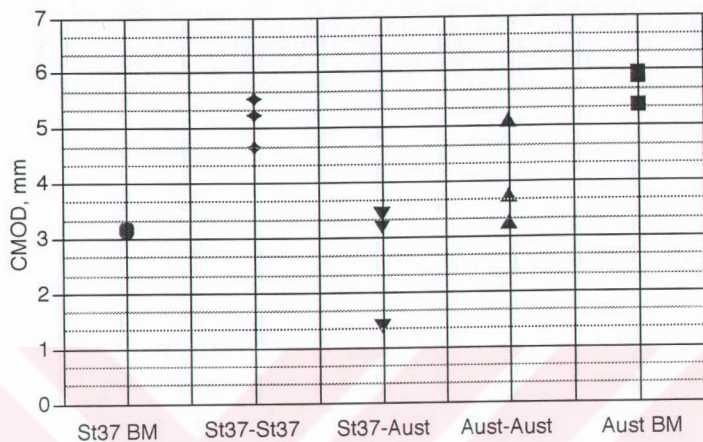


Fig. 4.41 a) CMOD vs. St37 BM, St37 similar, St37-Aust, Aust-Aust, Aust BM diagram at RT,

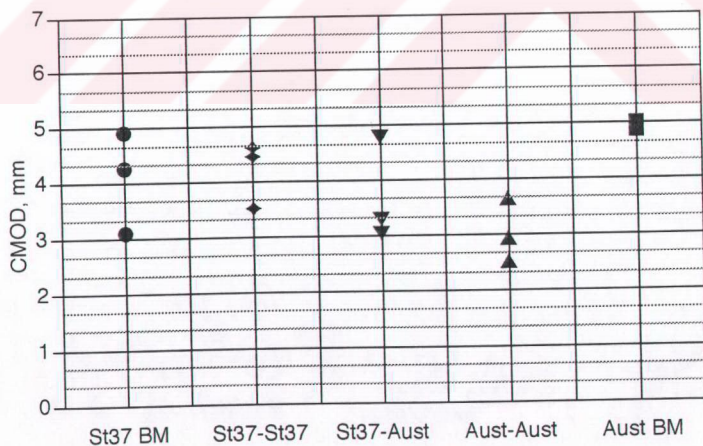


Fig. 4.41 b) CMOD vs. St37 BM, St37 similar, St37-Aust, Aust-Aust, Aust BM diagram at -40 °C,

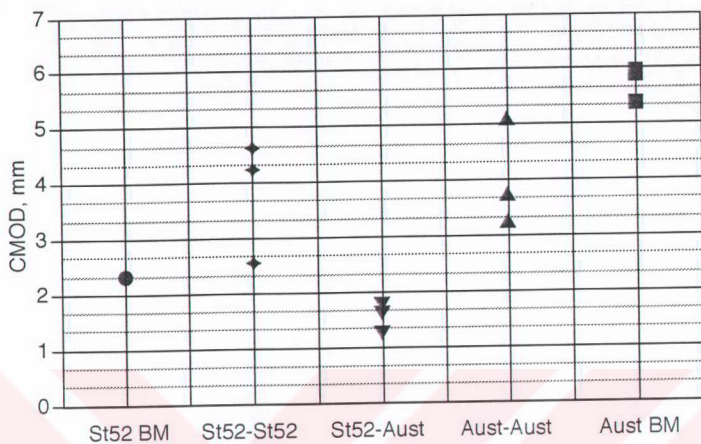


Fig. 4.41 c) CMOD vs. St52 BM, St52 similar, St52-Aust, Aust-Aust, Aust BM diagram at RT

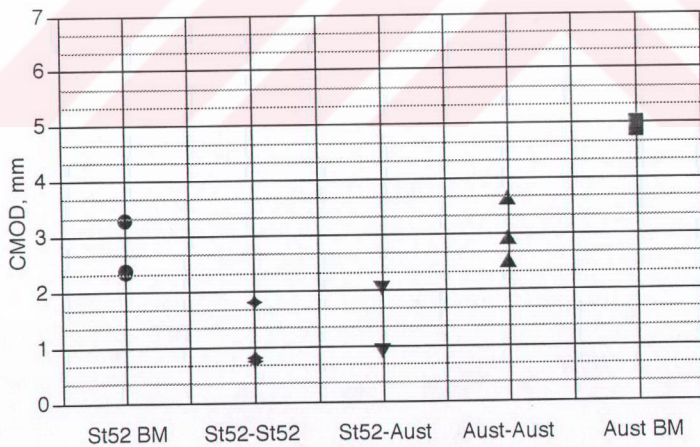


Fig. 4.41 d) CMOD vs. St52 BM, St52 similar, St52-Aust, Aust-Aust, Aust BM diagram at -40 °C

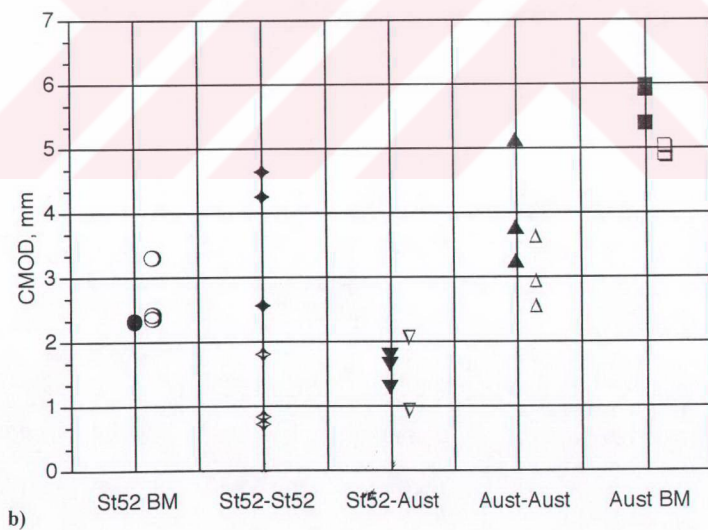
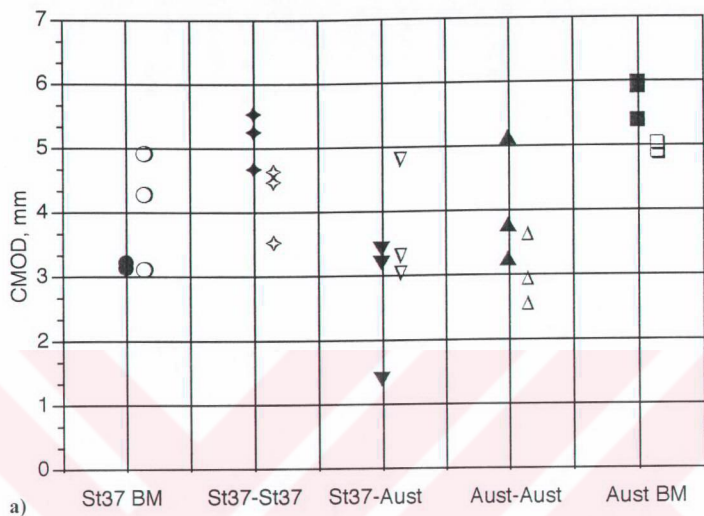


Fig. 4.42 Effect of temperature on CMOD values a) St37-aust joint b) St52-aust joint

CHAPTER FIVE

SUGGESTION FOR FUTURE WORK

Since the subject of strength mis-match of laser welds is a relatively new research area, this study comprises the initial steps in the understanding of strength mis-match effect in laser welded joints. A few areas uncovered in this study which could be a future work are;

- It was said that the existing correlation between hardness and strength of C-Mn steel weldments which were originally developed for conventional fusion welds may not be suitable for laser welds. A plot of the hardness vs. Yield strength (YS) values of the base metal, heat affected zone and weld metal parts of the specimens of the five weld combinations are presented in this study. But in order to determine the exact correlation between hardness and strength values of laser welded steel joints, further data is needed. This could be the subject of a future work.
- It is thought that carrying out longitudinal tensile and bending tests of laser welded specimens could be a suitable research subject in a future work since in service conditions longitudinal, as well as transverse deformation of a weld will occur in some situations. Therefore, it is relevant to test the weld in the longitudinal direction also. In longitudinal loading conditions, plastic deformation will be controlled by the dominating amount of base material and therefore, the deformation will be uniformly distributed along the weld.
- J integral calculations of the fracture toughness results at room temperature and - 40 °C tests are thought to be carried out using Ansys. The obtained results then, will be evaluated together with the CTOD results.

Various aspects of the similar and dissimilar joints of materials made by laser welding or other conventional material joining technologies still require a vast amount of R&D activities supported by the knowledge gained in long years of work unconventional materials. It is obvious that a better understanding of the microstructure-mechanical properties relationship of laser welded joints will feed-back the materials development activities both in conventional and new materials.



CHAPTER SIX

CONCLUSION

From the results of the literature review carried out on laser beam welding and strength mis-match of similar and bi-material joints and the tensile and fracture mechanics tests conducted at room temperature and -40°C , the following conclusions have been derived:

- Research carried out in laser welding has demonstrated that high quality welding can be performed for many ferrous alloys, nickel alloys and titanium alloys. This process has also been found effective for some iridium alloys and many dissimilar metals which are very difficult to weld by alternative processes. Some partial success has also been achieved for highly reflective materials such as aluminum alloys. Since laser welding is a highly robotized joining process, it requires more precise modeling and well defined characteristics compared to conventional welding processes. However, limited data is available on the laser beam welding of materials. The metallurgical aspects, mechanical properties and fracture behavior of laser welds and the response of conventional materials to this new process are also not yet fully established. Extensive research must be carried out in order to develop welding process parameters for laser welding of structural materials and to establish test standards for this type of welds.
- By their very nature, substantial differences in the microstructure and in strength properties (mis-matching) of the base, weld metal and HAZ may often occur in welded structures. Furthermore, the fracture behavior of a strength mis-matched joint is controlled by the local microstructure. Therefore, contributions from the base metal, weld metal and HAZ should be carefully analyzed in mis-matched joints

and relative test standards must be developed for such joints in order to take into consideration the effect of each region.

- The microstructures of CO₂ laser welded constructional C-Mn steels containing a ferritic constituent exhibit large proportions of bainite/martensite in the weld region due to rapid cooling involved. For dissimilar joints, solidification cracks parallel to the dendritic growth were observed within the weld zone.
- Extreme hardness increase in the HAZ and weld regions of ferritic similar joints and the HAZ-F and weld regions of ferritic-austenitic dissimilar joints were observed while there is no significant hardness increase in the weld region of the austenitic steel similar joints.
- All the transverse tensile specimens failed at the lower strength base metal sides due to the strength mis-match effect. Even the presence of some solidification cracks in dissimilar joints did not change the fracture location due to the strength overmatching of the weld region.
- The all weld metal flat microtensile specimens of the similar ferritic joint displayed significantly higher strength and lower strain values than the specimens extracted from the base metal due to the presence of bainite/martensite in the weld zone. The all HAZ specimens exhibited slightly higher strength values and lower strain values than those of the base metal.
- Flat microtensile specimens extracted from the base metal, HAZ and weld metal of similar austenitic steel joint exhibited similar strength and strain values as expected since no hardness peak was observed in these joints.
- The all weld metal specimens of the dissimilar joint exhibited high strength and extremely low strain values. The all HAZ specimens extracted from the HAZ at the

ferritic side also displayed higher strength and lower strain values than the ferritic base metal, while the specimens extracted from the HAZ at the austenitic side exhibited similar strength and strain values to those of the austenitic base metal.

- Successful application of the flat microtensile specimen technique to the laser welded joints, which exhibit very narrow weld regions has been demonstrated. By using this technique, it is possible to determine the local mechanical properties of the joints which can be correlated to the respective microstructure and hardness.
- All three point bend specimens failed in a fully ductile manner at room temperature tests. Pop-ins have been observed in St52 joints at -40 °C. At both temperatures, i.e. RT and -40 °C, extreme crack path deviation into the softer ferritic St37 base metal in similar St37 and dissimilar St37-austenitic joint was observed. In similar St52 joint, crack path deviation was also into the softer base metal. In dissimilar St52-austenitic joints cracks propagated along the border between the weld region and the austenitic base metal. For austenitic similar joints, the crack propagated within the weld region.
- There was no decrease in CTOD values in St52 and St37 base metals and St37-austenitic and St52-austenitic dissimilar joints at -40 °C tests. Decrease in CTOD values was observed in similar austenitic and ferritic joints and austenitic base metal specimens. Especially, a significant decrease is observed St52 similar joints compared to room temperature CTOD results. It is evident that St52 similar joint becomes brittle in -40 °C.
- The use of CTOD (δ_5) measurements for the (δ_5)-clip on a 5 mm gauge length perpendicular to the crack at its crack tip provides geometry independent CTOD-load curves for mis-matched bi-material panels. Therefore, the totally measured CTOD (δ_5) technique provides a unique methodology to measure fracture toughness values of bi-material joints.

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