

MECHANICAL ENGINEERING THEORY AND APPLICATIONS

A GUIDE TO
**LASER
WELDING**

ANTÓNIO MANUEL DE BASTOS PEREIRA
EVA SOLEDADE VAZ MARQUES
FRANCISCO JOSÉ GOMES DA SILVA

EDITORS

NOVA

It is my great pleasure to provide this endorsement for the new book from Nova Science Publishers, entitled *A Guide to Laser Welding*, a collection of the most recent contributions in the area of Laser Welding Technique, from both fundamental and applied perspectives, and edited by Professors António Pereira (Department of Mechanical Engineering, University of Aveiro), Eva Marques (PhD student), and Francisco Silva (School of Engineering, Polytechnic Institute of Porto), Portugal.

The book brings a total of 13 chapters, covering different aspects of the laser welding field. From laser welding equipment to numerical modeling and simulation, digital twins, and artificial intelligence, the final work provides a comprehensive and accessible review of the state of the art, also including metallurgy and quality assessment aspects in welded structures. Covering a wide span of themes within the laser welding research area, from the material science to the final products (including their performance on structural applications), passing by the fundamental aspects of mathematical and engineering characterization, product specification, limitations, and possibilities of different laser welding technique, the book can be a useful source of information for decision-making by experienced engineers, as well as a reference work for students seeking for beginning his/her research activities in the field.

Given the rapid and constantly evolving technology aspects in such a transversal (and integrative) area of research, the present book is definitely an added value for the future in the domain of laser welding techniques.

Professor Robertt Valente

*Head, Department of Mechanical Engineering
University of Aveiro, Portugal*

A Guide to Laser Welding

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**António Manuel de Bastos Pereira,
Eva Soledade Vaz Marques and
Francisco José Gomes da Silva**

A Guide to Laser Welding



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Preface

Laser welding is an advanced joining process that has followed the recent developments in other manufacturing processes, becoming an extremely used and investigated process. Several technologies developed in other areas of knowledge have converged in the area of welding, allowing significant improvements in the quality of the products obtained and in the increased productivity of the processes. Although some techniques have been implemented and studied for several decades, the interest of researchers in laser welding has not diminished, as it can be seen through new process techniques that have emerged or the welding of dissimilar materials, in addition to many other notable developments in laser welding, such as micro devices, simulation, and digital twins.

This work brings together a set of very interesting works, being a living proof that laser welding continues to be heavily investigated and that the developments around this manufacturing process and welding quality management systems are constantly emerging. Because the materials continue to evolve and the technology around welding also continues to develop at a very good pace, studies on the weldability of new materials and the application of new techniques and technologies to already well-known welding techniques do not stop happening.

The Editors are proud to have collected this set of works that can help scholars and researchers broaden their knowledge in the field of laser welding, thus contributing to the creation of a knowledge base that allows researchers to start new investigations and achievements in the coming decades.

Editors' Biographies

António Bastos Pereira did PhD and BSc in Mechanical Engineering at the University of Aveiro (Portugal). Founding shareholder and Director of Martifer Group (ca. 3500 employees) (1990–1999) – was responsible for the planning and production of about 500 steel structures and industrial equipment with a total amount exceeding 100 million euros. Since 2000 he has been a professor and researcher at the University of Aveiro, where he held management positions in the Department of Mechanical Engineering, as a member of the Executive Committee. He is currently the Director and Researcher at the Centre for Mechanical Technology and Automation (TEMA), where he coordinates more than 200 researchers. Recently, under his responsibility, TEMA received the highest rating and is included in the National Roadmap of Research Infrastructures of Strategic Relevance (Category 1 – those that have demonstrated high scientific potential and are considered to have high strategic regional and/or national relevance). TEMA was also selected, under his coordination, to appear on the European network of institutions considered to be relevant within the scope of KETs (Key Enabling Technologies), being one of the seven KET entities in Portugal. This network aims to help small- and medium-sized companies accelerate the process of commercializing their most innovative ideas, identifying partners of excellence in terms of research and technology transfer to society. KETs (advanced manufacturing technologies; advanced materials, industrial biotechnology, nanotechnology, etc.) are those considered to be innovative and with the potential to promote change in a variety of products and processes in traditional and emerging industries. Currently, Dr. António Pereira is also the Coordinator of the thematic line “Innovative and Sustainable Industries” of the largest associated laboratory in Portugal, with around 500 PhD researchers, LASI – Intelligent Systems Associate Laboratory. His main research areas focus on the study of materials and processes, from mechanics of composite materials to welding, collaborating on 34 research projects, and having been the Principal Investigator for 19 of them. António Pereira has more than 80 papers published in SCI journals, his h-index at Scopus is 26 with more than 1700 citations.

Francisco J. G. Silva did his PhD in Mechanical Engineering at Faculty of Engineering, University of Porto; MSc in Mechanical Engineering – Materials and Manufacturing Processes at Faculty of Engineering, University of Porto; BSc in Mechanical Engineering – Industrial Management at School of Engineering, Polytechnic of Porto, Portugal. He has also a Post-Graduation in Materials and Manufacturing Processes at Faculty of Engineering, University of Porto, Portugal. He was Head of the Master's Degree in Mechanical Engineering at School of Engineering, Polytechnic of Porto, Portugal, the position was occupied from 2014 to 2022. He also was Head of the Bachelor's Degree in Mechanical Engineering at the Superior School of Industrial Studies and Management, Polytechnic of Porto, from 2003 to 2006. He has supervised some PhD students at Faculty of Engineering, the University of Porto, as well as more than 190 students at School of Engineering, Polytechnic of Porto. Moreover, he co-supervised more than 75 MSc students at School of Engineering, Polytechnic of Porto, and Faculty of Engineering, University of Porto. He has more than 220 papers published in peer-reviewed journals with impact factor and more than 125 presentations at International Conferences. He is a member of the Editorial Board of the scientific journal *Coatings* (MDPI), *Machines* (MDPI), *Materials* (MDPI), *Metals* (MDPI), and *Solids* (MDPI). He has reviewed more than 500 papers for reputed journals with high impact factor, such as *Journal of Cleaner Production*, *Composites Part A*, *Wear*, *Surface and Coatings Technology*, *Metals*, *Materials*, *Coatings*, *Journal of Materials – Design and Applications*, amongst many others. He is a member of the Scientific Committee of some International Conferences, such as FAIM – Flexible Automation and Intelligent Manufacturing, among others. He has conducted some Special Issues in MDPI journals, such as *Coatings*, *Materials*, *Machines*, *Sustainability* and *Metals*. Moreover, he has conducted some national and international research projects involving other academic institutions and companies. He has a strong linkage to the industry, due to his past link to the Industrial Sector, having been the founder, owner, and manager of a company in the electric sector for 18 years.

Eva S.V. Marques is a PhD student of Mechanical Engineering in University of Aveiro, currently researching laser welding processes at the Centre for Mechanical Technology and Automation (TEMA), already with published papers in *Metals* (MPDI) and *Materials* (MPDI), MSc in Mechanical Engineering – Materials and Manufacturing Processes at School of Engineering, Polytechnic of Porto, and a BSc in Mechanical Engineering in Superior School of Industrial Studies and Management, Polytechnic of Porto, Portugal. She has also a post-graduate specialization in Lean Management by CLT-Services. Still working presently in the industry, she has already been through different areas

such as controlling (WEG), logistics, and continuous improvement (RACE-SONAE CAPITAL), where she was accountable for the implementation of a continuous improvement model in the company and production/project planning and management. Subsequently, she was also a Process Engineer at CAETANOBUS, where she was responsible for the development of the mass production of the Hydrogen Fuel Cell Bus, in partnership with TOYOTA CAETANO PORTUGAL. She is now linked to technological and digital areas, currently working as a Quality Assurance Specialist at FARFETCH.

Chapter 1

Lasers for Welding

A. B. Pereira^{1,2}, F. A. O. Fernandes^{1,2},

Eva S. V. Marques¹, and Francisco J. G. Silva^{1,3,4}

¹TEMA – Centre for Mechanical Technology and Automation, Department of Mechanical Engineering, University of Aveiro, Campus de Santiago, 3810-193 Aveiro, Portugal

²LASI – Intelligent Systems Associate Laboratory, 4800-058 Guimarães, Portugal

³INEGI – Instituto de Ciência e Inovação em Engenharia Mecânica e Engenharia Industrial, Rua Dr. Roberto Frias 400, 4200-465 Porto, Portugal

⁴ISEP – School of Engineering, Polytechnic of Porto, Rua Dr. António Bernardino de Almeida 431, 4200-072 Porto, Portugal

Abstract

This work presents a brief study of lasers that can be used in welding. Initially, an analysis is done on the properties of light and its differences with monochromatic, coherent, and directional laser light. The main types of lasers on the market are those with a solid active medium, such as Nd:YAG or fiber, and those with a gaseous active medium, such as CO₂ lasers. Currently, with the increasing ease of electronic process management, solid-state lasers reach ground that was previously exclusive to CO₂ lasers, namely in welds with high penetration.

Keywords: Laser welding; laser modes; laser types

Introduction

The Sun is still the main source of energy through which our planet survives. In fact, humanity has been looking for multiple energy sources, but it is certain that without natural light, life as we know it would be impossible. Light is easily absorbed by multiple materials, especially natural ones, and this interaction

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with the molecules that constitute them causes them to vibrate and generate heat.

Isaac Newton already knew that sunlight can be decomposed into a set of colors called the continuous spectrum. In fact, each of the colors in this spectrum corresponds to electromagnetic waves with different wavelengths. The distance between two consecutive peaks, λ , as shown in Figure 1.1 is called wavelength.

The electromagnetic spectrum is continuous; however, it is customary to assign different designations to certain wavelength ranges. For example, visible light is only between the wavelengths of 0.4×10^{-6} m and 0.7×10^{-6} m ("Vis" zone in Figure 1.2).

Radiation transports energy, and the longer the wavelength, the shorter the energy and vice versa. The relationship between the absorbed energy, E , and the frequency, f , is given by the Max Planck equation. In this equation, it is assumed that the photons of light, that is, the elementary particles (also known as quantum) are transferred to the molecules and atoms of the irradiated bodies, and that, therefore, this increase in energy is directly proportional to the frequency of the photon (Equation 1.1).

$$E = h \cdot f \quad (\text{Equation 1.1})$$

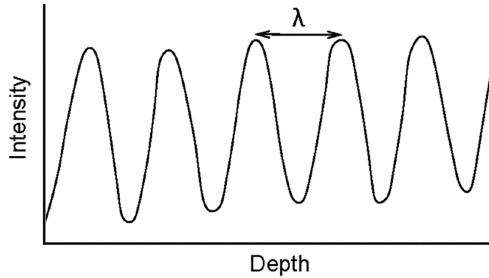


Figure 1.1 Wavelength (Mohammadi et al. 2019).

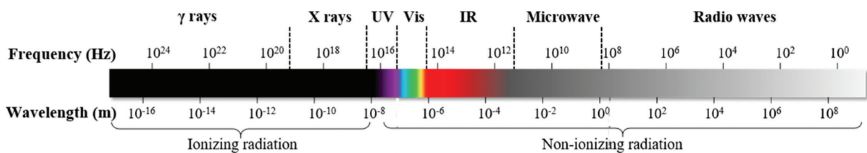


Figure 1.2 Electromagnetic spectrum (Zhang et al. 2019).

where $h = 6.626 \times 10^{-34}$ J.s is Planck's constant, and the frequency is given by Equation 1.2.

$$\lambda = \frac{c}{f} \quad (\text{Equation 1.2})$$

being c the speed of light, that is, 300,000 km/s.

It is the fact that light converts into heat (energy) when absorbed, the possibility of it being able to travel enormous distances without great losses, as well as being later concentrated in a point, leading to a significant increase in the power density (W/mm^2), which culminated in lasers.

The Laser

Isaac Newton described, in 1704, that light can be considered a stream of particles. Later, in 1803, Young, with the discovery of the polarity of light, identified its wave behavior. In 1905, Albert Einstein proposed the concept of quanta, in which he argues that light is composed of individual particles and that these have a discrete amount of energy. Today, these particles are called photons. Maxwell further defended light as a rapid vibration of an electromagnetic field resulting from oscillations of particles. In the past, when the phenomenon of blackbody radiation was discovered at the beginning of the 20th century, Maxwell's theory was discarded. This theory, published by Max Planck in 1900, held that light was a form of electromagnetic radiation. Without this principle, the laser would never have been invented.

Einstein's great contribution to laser technology came in 1917 when he discovered that if a population inversion between higher and lower energy levels were possible, it would be possible to obtain amplified stimulated radiation. However, this principle was only used for more practical purposes in the 1950s, when mechanisms for harnessing laser energy began to be developed (Bromberg, 1991).

At this stage, work was being done on the creation of what was called MASER (Microwave Amplification by the Stimulated Emission of Radiation), a device capable of amplifying microwaves for application in microwave communication systems. At the time, it was believed that it was possible to create an optical maser, a device capable of creating powerful beams of light using high-frequency energy (Smith, 1957). It was Theodore Maiman who in 1960 invented the first laser with the name "LOSER" (Light Oscillation by

Stimulated Emission of Radiation) that worked in a pulsed way and emitted a red light (Mainan, 2000).

The laser was a remarkable technological advance, although in its early years it was a technology that served no purpose. It was not potent enough to be used in weapons for the war industry, and its usefulness for transmitting information through the atmosphere was severely impaired due to its inability to penetrate clouds and rain (Bromberg, 1991).

However, the evolution continues to this day. For example, laser is used for high-precision eye surgery and for most diverse applications in the industry whether in low or high power, from marking parts to cutting or welding metals.

Laser Properties

Laser is the acronym for Light Amplification by Stimulated Emission of Radiation.

What differentiates lasers from other types of light sources, such as the Sun, is the way the light is created.

Unlike sunlight, which is radiated in all directions, with all light wavelengths and out of phase with each other (Figure 1.3), the laser beam is (Figure 1.4) (Dawes, 1992):

- Monochromatic – the wavelength is the same for all waves
- Coherent – all waves are in phase
- Directional – all waves are approximately parallel

This “directional” property is of special importance for material processing. Since the light waves all have the same direction, it is possible to create a compact beam of light that contains a lot of power. This beam can be conducted

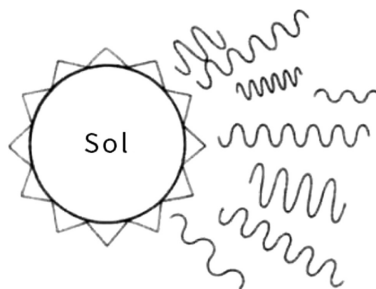


Figure 1.3 Light waves from sunlight (Dawes, 1992).

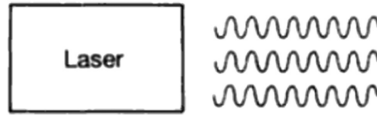


Figure 1.4 Light waves from the laser (Dawes, 1992).

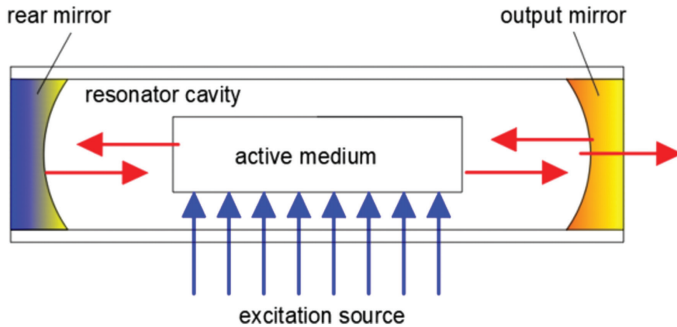


Figure 1.5 Composition of the laser system.

over long distances and can be applied to materials with a very small diameter focus dramatically increasing the power density applied at that location. The power density (W/mm^2) is 10^7 times higher in the laser beam than in an incandescent lamp (Dawes, 1992).

Composition of the Laser System

Although there are different types of laser, they all obey the same physical principles to generate the laser beam. Figure 1.5 shows the schematic of the main constituents responsible for originating the laser beam.

The resonant cavity consists of two mirrors with different reflectivity indices placed parallel to each other to form an optical oscillator that functions as a cavity through which the light will travel.

Between the mirrors is the active medium, which is the substance through which the photons oscillate and the laser is amplified. This substance can be a gas, a liquid or a solid, which, when excited, emit radiation in the form of light waves (photons).

In order for the active medium to emit light, it is necessary to supply energy in order to excite it. The way in which energy is supplied depends on the active medium used. However, energy is normally provided by optical, electrical, or chemical processes.

One of the mirrors is completely waterproof and reflects all photons. The other is partially transparent and allows part of the photons to pass through it forming the laser beam.

A large part of the energy is not converted into laser light, but into heat. Depending on the laser power, there is a cooling system, usually air or water, which dissipates the heat to ensure that the active medium and the resonant cavity do not overheat, guaranteeing the correct operation of the laser.

Operating Principle

Laser technology has evolved considerably, and today there are several types of laser with different applications and purposes. However, the operating principle is similar for the various types of laser available.

To understand the working principle of a laser, it is really important to review a little atomic physics. An atom consists of a positively charged central nucleus surrounded by a cloud of negatively charged electrons. The atomic nucleus is composed of protons and neutrons. Figure 1.6 shows a schematic representation of the structure of the atom.

When energy is supplied to an atom, electrons seek new positions to be in equilibrium. This may correspond to moving to a higher energy orbit. Upon arrival at that point, the electron tries to return to its ground state. It is at this point that photons are released, which are nothing more than the energy absorbed by the electrons discharged in the form of photons.

The creation of a laser beam essentially consists of three steps that occur almost instantly, which are as follows:

- I. The source supplies energy to the active medium, exciting the atoms in such a way that electrons are temporarily raised to higher energy states. As the electrons cannot remain there indefinitely, they are forced to descend to a lower energy level. In this situation, the electron loses the energy obtained with the energy of the source through the emission of a photon. This is called spontaneous emission, and the photons produced by this method are the seeds for generating the laser beam.
- II. Photons emitted by spontaneous emission can eventually collide with electrons that are in other higher energy states. The photon colliding with the electron causes it to descend to a lower energy level giving rise to a new photon. These two photons are coherent, meaning they are in phase, have the same wavelength, and travel in the same direction. This phenomenon is called stimulated emission.

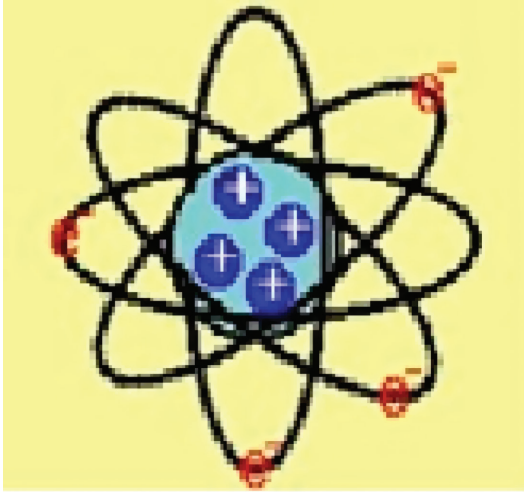


Figure 1.6 Schematic representation of the structure of the atom (Aysegül, 2019).

III. Photons are emitted in all directions causing them to travel through the active medium until they collide with the mirrors of the resonant cavity and are reflected back to the active medium. For amplification to occur, there must be more atoms in the excited state than atoms at lower energy. This is called “population inversion” and is mandatory for the generation of laser light.

Types of Laser

With the constant evolution that laser technology has undergone, it is possible to find different types of laser that serve the purposes of the most different areas of application. Their designation is based on the material that constitutes the active medium of the equipment, thus existing three categories: solid-, liquid-, and gaseous-state lasers. Solid-state lasers can be further subdivided into categories according to the geometry of the active medium. These can be rods (rod lasers), disks (disk lasers), and fibers (fiber lasers).

Figure 1.7 represents a diagram of the classification of the different principal types of laser.

Different lasers have substantially different wavelengths. In fact, they range from 200 nm to 10,000 nm, so the way they will be absorbed by materials during welding or cutting processing is completely different.

Dye lasers are not used in welding due to their generally high price. Its widest range of applications is in medicine or research.

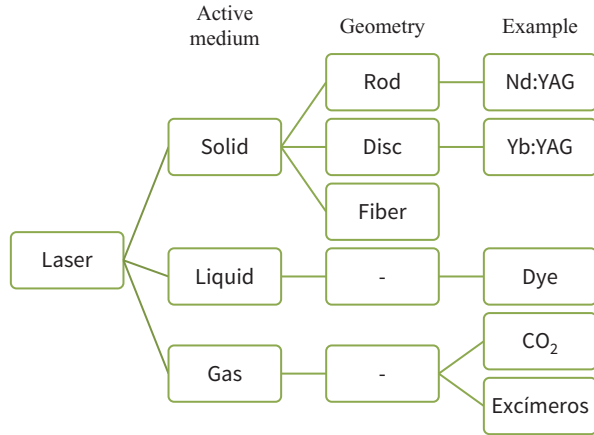


Figure 1.7 Types of laser.

In fact, there are several types of solid-state lasers. These lasers differ in the shape and material of the active medium, in the design of the resonant cavity, and in the type of excitation source. This diversity is what gives lasers so much versatility. These lasers are commonly used in welding, cutting, and drilling.

Solid-State Lasers

Solid-state lasers are usual in industrial production. The first lasers were characterized by having short laser pulses with a power near to 1 kW, which gave them good characteristics for spot welding.

Today there are many different solid-state lasers. These lasers differ in the shape and material of the active medium, in the design of the resonant cavity, and in the type of excitation source. This diversity is what gives lasers so much versatility. These lasers are commonly used in welding, cutting, and drilling (Werkzeugmaschinen, 2007).

Cylinder Laser

The Nd:YAG (Yttrium Aluminum Garnet – $Y_3Al_5O_{12}$) lasers available on the market have powers between 100 W and 1000 W. Despite this wide range of powers, these lasers have an even lower power than CO_2 lasers, since they can reach peak powers in the order of 10 kW. In terms of performance, a 150 W Nd:YAG laser is able to weld a 0.5-mm-thick plate at a speed of 0.3 m/min, while a 1 kW laser can easily weld a 4-mm-thick plate at the same speed (Dawes, 1992).

The most important aspect of this type of laser is the wavelength of the beam ($1.06\text{ }\mu\text{m}$) that can be transmitted through a fiber optic cable, which makes this laser very attractive for high-speed welding production and automation.

The schematic representation of Nd:YAG lasers is illustrated in Figure 1.8.

The Nd:YAG laser uses a solid crystalline active medium consisting of a YAG cylinder. The YAG cylinder is located in the resonant cavity being excited by flash lamps (krypton or xenon). The number of lamps to be used essentially depends on the maximum power to be emitted. These lamps can also be arranged in various ways.

Laser radiation is produced through stimulated emission. This has to be amplified in a resonant cavity to be later emitted. Laser light is then produced in the Nd:YAG cylinder. Its dimensions are important as the larger the cylinder, the greater the maximum power available to the laser.

Fiber Laser

The main characteristics of fiber laser are high efficiency and small dimensions. It is produced with double-layer optical fiber made of silica. For the purpose of signal amplification, the lower part is doped with different types of ions, such as ytterbium. In turn, the outer layer propagates the excitation light

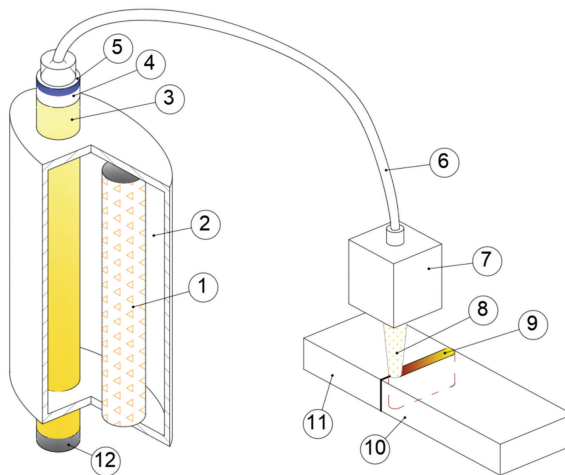


Figure 1.8 Schematic of the Nd:YAG laser welding process: (1) flash lamp; (2) resonant cavity (elliptical mirror); (3) Nd:YAG cylinder; (4) output mirror; (5) focusing lens; (6) optical fiber; (7) welding head (supported by drive system); (8) laser beam; (9) welding bead; (10) and (11) parts to be welded; (12) reflective mirror.

so that it is absorbed by the core and thus propagation is possible, and this layer is made of glass or polymeric materials with reduced refractive indices.

The main advantages are the high performance achieved, low maintenance, good beam quality, and the obtained focal point is already smaller. As a disadvantage, it presents the high cost of the equipment; however, as the need for maintenance is low, it ends up offsetting the initial investment.

Figure 1.9 presents the working scheme of a fiber laser.

Diode Laser

This type of laser is distinguished by being compact and its components are semiconductor. The source of energy is electricity, having no need for any other source of excitation. As seen in the previous point, they can also be used as a source of excitation. Diodes are usually formed by combining elements from Group III and IV of the periodic table.

These materials have a conduction band and a valence band. First, the valence band is completely occupied and the conduction band is completely free. When the valence band is excited, some electrons shift to the conduction band and others go to lower energy levels within the valence band. New layers of the bands are being formed and the process is repeated. Figure 1.10 shows a schematic of the basic operation principle for a diode laser. The flow of electric current is between the two semiconductors: one of type P and the other of type N. A coherent radiation will appear between them, that is, the light waves generated by the device have the same frequency and phase

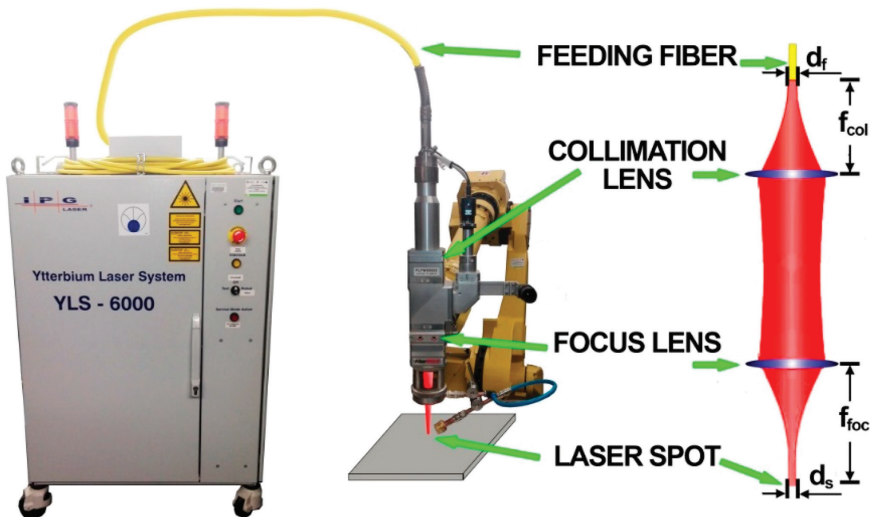


Figure 1.9 Fiber laser operation (Landowski, 2020).

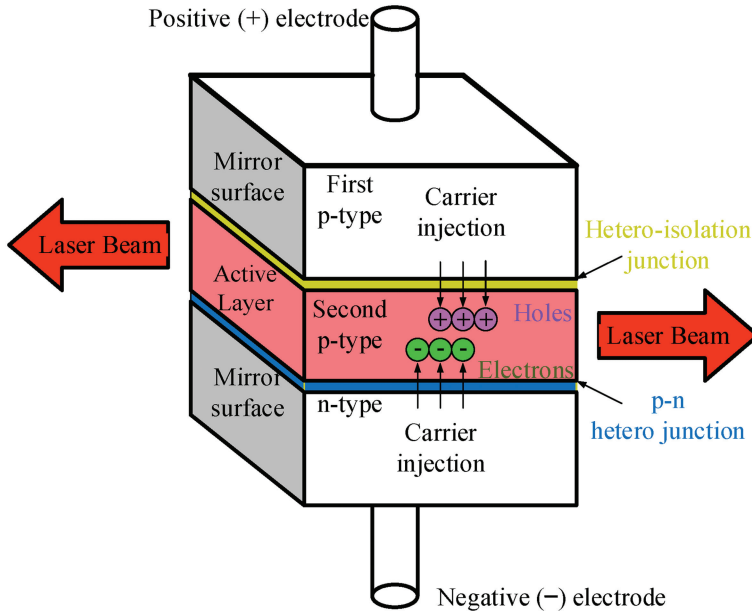


Figure 1.10 Diode laser operation (Pai et al., 2021).

The main disadvantage of this type of laser is the relatively large diameter of the focal point, which translates into a low energy density. However, it is possible to join several individual diode outputs in an optical fiber and thus increase the energy density required for soldering. Even so, this type of laser is mainly used for welding thin sheets.

Lasers in the Gaseous State

The most used gaseous lasers are excimer and CO₂ lasers.

Excimer Laser

The active medium consists of rare gases, helium, or halogen. It is well known that noble gases do not tend to join with other elements; however, when they are excited by an electrical discharge or a beam of electrons, their atoms become ionized. These ions can bind to chlorine or fluorine molecules forming excited dimers. This bond is quite strong, but short-lived. When the bond is broken, energy is released in the form of photons. Thus, this laser produces short pulses, but of high power.

The high quality of the result, due to extreme precision, is the main highlighted advantage.

CO₂ Laser

The active medium consists of carbon dioxide (6%), which emits radiation, nitrogen (84%), which helps in the excitation of CO₂, and helium (10%), which promotes cooling. This laser system is characterized by the following parameters: internal optical path, excitation mode, and cooling type.

The excitation of the active medium can be by radiofrequency or direct current. Radiofrequency enables excitation in pulsed mode. Direct current excitation requires two electrodes where an electrical discharge is applied.

The optical path is said to be optically stable when the beam passes through only one mirror. On the other hand, it will be optically unstable if the resonant cavity has 100% reflective mirrors and the beam is reflected to the outside.

Cooling can be done by diffusion, transverse flow, fast axial flow, and slow flow.

Areas of Application

Laser machines have revolutionized the industry due to their attractive advantages. These machines have the following characteristics, among others:

- Non-contact: the lasers do not mechanically touch the workpiece
- High precision: lasers can reproduce contours with high precision
- Local heating: the thermally affected area is very small
- Flexibility: a huge variety of shapes and contours can be made with a single tool
- Accessibility: access from only one side of the joint, which is a huge advantage when compared to resistance welding

Despite the advantages of laser technology, it is important to weigh its benefits against the investment required when compared to other alternative technologies. However, the following should be noted:

- Lasers are fast. The processing speed of laser technology is generally higher than alternative technologies
- Lasers guarantee quality. The perfection of laser work is usually better than that of conventional technologies
- Lasers reduce finishing work. The lasers allow, from the outset, a good finish and further improvement steps are unnecessary

All these characteristics make the production speed increase, increasing productivity and compensating for the investment in laser technologies. It is these characteristics that make the laser competitive with conventional technologies.

Laser Welding Modes

When the laser beam is irradiated on the surface of the part, the energy absorbed by the material causes its fusion.

The flexibility of the laser offers three types of welding: by conduction, conduction/penetration and penetration, or keyhole. Figure 1.11 represents the extremes types of welding.

Figure 1.11 (left) shows the conduction mode. This welding mode is characterized by a low power density (10^4 – 10^6 W.cm⁻²), thus resulting in a bead with little penetration. The laser beam melts the material along the bead and when it solidifies, it forms the weld bead. This mode is generally used to weld thin parts, Figure 1.12.

Figure 1.11 (right) shows the penetration or keyhole mode. This welding mode is defined by large weld penetrations and requires high-power densities (over 10^6 W.cm⁻²). In this process, the laser beam hits the part and quickly raises the temperature of the part to the melting point. As the material starts to melt, it also starts to release steam. The vapor that is dissipated exerts pressure on the molten material, keeping the cavity open during welding. Thus, the laser is able to reach the fusion zone of the part in deeper areas, resulting in higher penetration welds. The result is a deep, narrow, steam-filled hole called a keyhole. The weld depth can be ten times greater than the bead width. This welding mode is characteristic of high-power continuous lasers or high-energy pulsed lasers (Figure 1.13).

Between these two energetic extremes, there is a transition or mixed conduction/penetration mode. This case occurs when the power density is around the critical point of 10^6 W.cm⁻². The result is a bead with greater penetration than in the conduction mode, but not having keyhole.

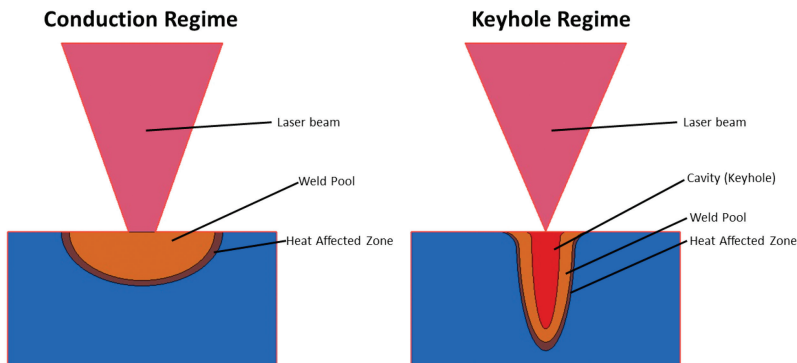


Figure 1.11 Schematic representation of the energetic extreme types of welding (Sánchez-Amaya, et al., 2017).

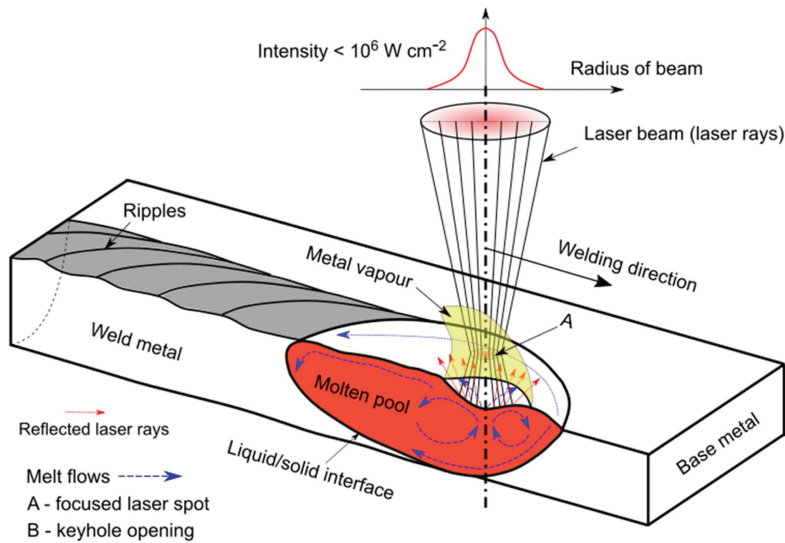


Figure 1.12 Schematic representation of the conduction welding principle (Bunaziv, 2021).

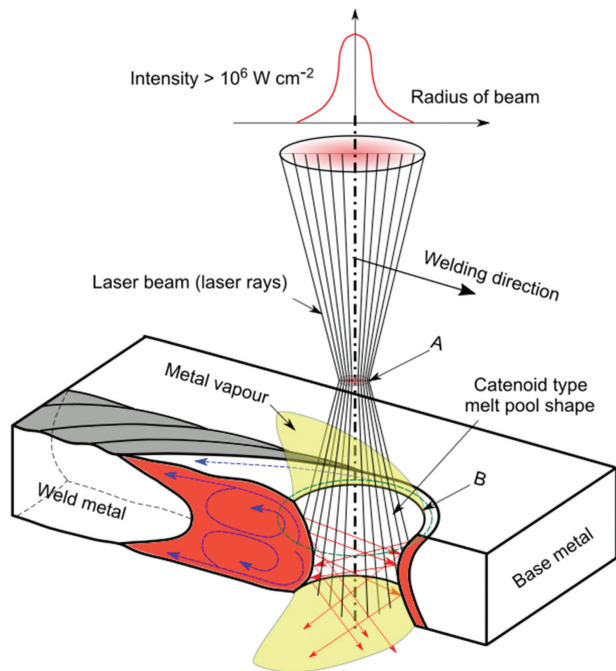


Figure 1.13 Schematic representation of the penetration welding principle (Bunaziv, 2021).

Conclusion

With this study, the following conclusions can be drawn:

- There is an appreciable range of lasers that can be used in welding operations, either with solid active media, such as Nd:YAG, or with gas, for example, CO₂. Liquid-state lasers have no application in welding practice.
- Currently laser continues to revolutionize welding, especially with the updating of solid-state processes and its simpler manipulation by electronics. In addition, the conduction through a flexible optical fiber greatly facilitates the robotization of the welding process, with, however, the inherent dangers of radiation falling outside the product to be worked, which requires extra care.

Funding

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Chapter 2

A Review of Laser Weld Formation and Beam–Material Interactions

T. Patterson¹

¹INL – Idaho National Laboratory, Energy and Environment Science and Technology, Idaho Falls, Idaho, United States of America

Abstract

Laser beam welds are generally categorized as resulting in two distinct modes, known as conduction and keyhole. Conduction mode welds are relatively wide and shallow with a low height-to-width aspect ratio, and keyhole mode welds are narrow and deep with a high height-to-width aspect ratio. The difference between these regions is a combinatory result between process parameters and material properties. High power and/or small beam diameters result in high power densities that enable metal vapor to impart a back pressure into the molten weld pool. It is this vaporization and the ability to couple the energy deeper into the workpiece that results in deep penetration. The energy absorption into the material is based on the photon wavelength, angle of beam incidence, power density, atmospheric conditions, plume formation, and material properties. These physical beam–matter interactions change the overall shape of the weld pool and influence defect formation such as porosity, root spiking, and/or solidification cracking. In addition, the solidification rate within the weld pool is strongly influenced by the weld pool geometry and changes the resultant microstructure. The transition between conduction and keyhole is not distinct and varies based on the laser parameters, material composition, shielding gas, and atmospheric pressure. This chapter elucidates the parameters that develop distinct weld pool geometries and the physics associated with laser beam welding and weld pool development.

Keywords: Laser beam welding; Weld pool formation; Conduction mode; Keyhole mode; Vaporization; Solidification

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Introduction

Lasers have become a major welding process used in industries requiring high precision or deep penetration welding. The advancements in laser technology have greatly expanded the use of lasers for material joining. Particularly, high-power solid-state, fiber, and diode lasers are becoming increasingly competitive to other fusion welding processes such as gas tungsten arc welding (GTAW), electron beam welding (EBW), and resistance welding. Historically, high-power welding lasers used bulky focusing optics attached to mechanized or automated positioning systems. However, recent developments (c. 2020) have resulted in high-power (2.5 kW) handheld laser welding equipment.

The advantages of laser beam welding (LBW) are largely based on the high energy density, which results in a narrow heat-affected zone and less component distortion. A benefit LBW has over EBW, a competitive high energy density process, is the ability to weld under atmospheric conditions. This is attributed to reduced beam attenuation based on the nature of photons versus electrons. However, lasers have higher reflectivity, which can result in challenges when welding reflective metals. Many other physical phenomena associated with LBW result in unique welding capabilities and weld formation. LBW is an easily automated process that creates reduced distortion and may be capable of performing dissimilar metal welds. The range of weld profiles, physical interactions, and resulting weld defects will be further discussed.

Weld Profiles

To produce a weld, sufficient laser beam energy must be absorbed into the workpiece to overcome conductive heat loss, latent heat, and radiation losses. These losses and energy absorption are dependent on the laser wavelength, power, beam diameter, energy distribution, travel speed, material composition, and environmental conditions. High energy density beam processes were historically regarded as developing two distinct weld profiles, known as conduction and keyhole mode welding. Conduction and keyhole mode welding are based on quantifying a discrete transition from a low depth-to-width aspect ratio with shallow penetration to a high aspect ratio, deep penetration weld. An example of conduction versus keyhole mode weld profiles is shown in Figure 2.1.

The laser welds shown in Figure 2.1 were autogenous welds, i.e., no filler metal, deposited on Ti-6Al-4V. The conduction mode weld shows a depth-to-width aspect ratio of approximately 1, whereas the keyhole mode weld shows

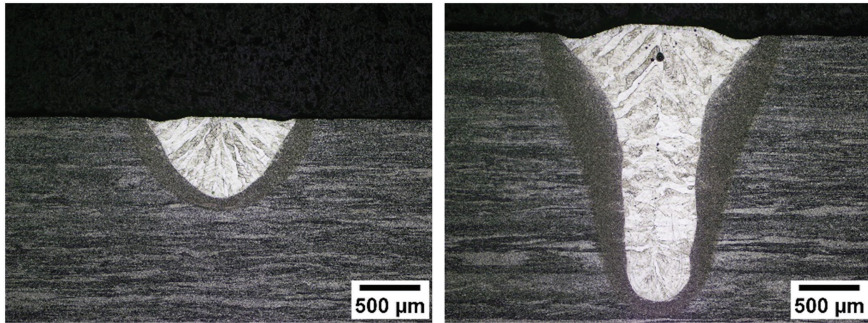


Figure 2.1 Comparison of conduction mode (left) and keyhole mode (right) for laser welds in Ti-6Al-4V (Tate Patterson, 2021).

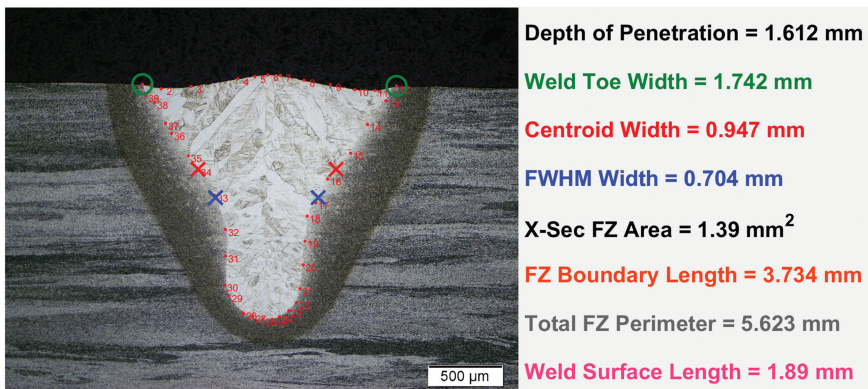


Figure 2.2 Measurements of a fusion zone cross-section used in neural network predictions of laser weld profiles (Tate Patterson, 2021).

a depth-to-width aspect ratio of approximately 5. The capability to generate welds with drastically different profiles is one benefit that LBW has over arc-based fusion welding processes. Typically, the use of conduction or keyhole mode welding is based on the application and penetration requirements, and not determined based on the material being welded.

Depth-to-width ratios are a loose representation of weld characteristics because the weld width can change drastically throughout the weld depth. As shown in Figure 2.2, the fusion zone width at half the penetration depth is less than half the weld toe width. This value is denoted in the figure as full width at half maximum (FWHM). Welds with low penetration, i.e., conduction mode, will have weld toe widths that more closely approximate FWHM measurements or centroid-based weld widths. The centroid width is measured from

the fusion zone's transverse cross section. Laser beam welds performed in the atmosphere will likely result in welds with wider nail head (surface) shapes and narrow weld roots or fingers. However, welds made under reduced pressure or power densities high enough to approach laser cutting (10^6 W/mm²) result in narrow weld toes, which are only slightly larger in width than the remaining depth of the weld. This is described in more detail in the "Keyhole Mode" section.

Conduction Mode

Welds produced in conduction mode are a result of a low-power density and/or a short interaction time, that is, a high travel speed. These weld profiles can be produced with continuous-wave or pulsed-wave lasers, but it is generally considered that no or minimal vaporization develops. The description of conduction mode weld is based on the resulting thermal diffusion or conductive heat transfer into the underlying material. Within this regime, penetration and weld widths are dictated by the material's thermal diffusivity relative to the viscosity of the molten weld pool (Chan et al., 1988; Webber et al., 2011). Higher thermal conductivities generally result in spherical weld pool geometries with minimal influence from surface tension-driven fluid flow. When the weld pool viscosity is high and the molten metal flows are dominated by surface tension, the weld shape results in shallow and wide weld pools.

Conduction mode welding is generally used in applications requiring low penetration and reduced heat input. Components commonly conduction mode welded include electronics, thermal batteries, pacemakers, stents, jewelry, etc. These applications require precision, minimal heat input, no/limited distortion, and/or hermetically sealed joints. Conduction mode welds can be either spot welds or linear welds and are commonly performed autogenously, but filler metal can be added.

Aluminum is an example of a metal alloy that is easier to weld within the conduction mode regime. Keyhole welds in aluminum alloys are prone to porosity entrapment and solidification cracking in the fusion zone (Eurico Assuncao & Williams, 2014). Polymers are also joined with conduction mode welds and can be joined using CO₂ lasers with a wavelength of 10.6 μ m. The long wavelength of CO₂ lasers lies within the absorption band of most polymers. If the laser wavelength is transparent to the polymer, for example, Nd:YAG, fiber, and diode lasers, filler, coloring agents, or coatings must be applied to the joint to absorb the energy and provide localized melting, or one of the polymers must be capable of absorbing the beam energy. The process where one polymer is transparent to the laser and backed with a material

that absorbs the laser energy is known as laser transmission welding (LTW) (Grewell et al., 2003). A practical example of LTW applications is joining the housings on automotive lights.

LBW in conduction mode is also used for additive manufacturing to selectively deposit metal powders in powder bed fusion (PBF) or powder directed energy deposition, also known as laser-engineered net shape (LENSTM) (Atwood et al., 1998). These processes are better suited for minimal melting of the deposited material without penetration into the underlying layer. Low-power density conduction welds allow for better process control and improved weld quality. Vaporization is also reduced to minimize elemental loss and mitigate pore formation. However, due to the small beam sizes ($\approx 100\ \mu\text{m}$) used in laser PBF additive manufacturing, it is likely that these welds develop vapor cavities within the melt pool (Cunningham et al., 2019).

Defects

Conduction mode welds can be susceptible to solidification cracking. The susceptibility is a result of a rapid solidification rate, alloy composition, impurity content, and internal shrinkage stress. Because conduction mode welds maintain low levels of vaporization, weld pool porosity is generally not an issue. However, this is material dependent and may be exacerbated with pulsed LBW. Other issues may include unfavorable phase fractions caused by rapid solidification and cooling rates. Structures with these issues include steel and titanium alloys prone to forming fully martensitic structures (T. Patterson et al., 2022) or stainless steels and duplex stainless steels that form insufficient ferrite or austenite fractions, respectively (Tate Patterson et al., 2022).

Keyhole Mode

Keyhole mode is a term used to describe the cross-sectional shape of deep penetration welds that result from melting and vaporization. The power density is high enough and the travel speed is low enough that conduction, convection, and radiation heat loss are inadequate at preventing vaporization. The vaporized metal, and potentially plasma, generates a back pressure onto the weld pool, and this pressure results in the depression forming a vapor cavity (Cho et al., 2009; Lee et al., 2002). Once this cavity is formed in the molten pool, higher energy is absorbed by the material because the photons become effectively trapped within the vapor cavity or keyhole. Ideally, there is a constant vapor pressure exerting an equilibrium force on the surrounding fluid. However, the complex nature of the solid–fluid–vapor–beam interaction governing the established keyhole is far from an equilibrium process (Kroos

et al., 1993). The nature of beam–material interactions induces transient events resulting in keyhole instability (Matsunawa, 2002). These events may lead to root-spiking defects and/or keyhole collapse issues in deep penetration welds (Elmer et al., 2022). Pores resulting from the collapse of the vapor cavity are usually trapped within the root and cannot implode or float out under rapid solidification. Other pores may develop from absorbed atmospheric air and/or shielding gas that also cannot float out prior to solidification.

Keyhole mode welding is used in applications requiring weld penetration deeper than what is achievable via conduction mode welding. The high energy density allows for fast travel speeds and increased productivity compared to arc welding processes. Additionally, the low heat input reduces residual stress formation and overall component distortion. Issues with keyhole mode welding are generally pore formation due to keyhole collapse and entrapped atmospheric air or shielding gas. Keyhole mode welds can also be susceptible to solidification cracking caused by high travel speeds resulting in rapid solidification rates and susceptible weld pool geometries (Tate Patterson et al., 2022).

Applications of keyhole mode laser welding include joining of automotive drive trains, gears, appliances, exhaust system components, automotive door panels, etc. The deep penetration welds are commonly performed autogenously, but filler metal can be added depending on the application. Filler is usually added via wire feed or filler metal shims preplaced in the joint prior to welding. The relatively small width of millimeters or less cause difficulty in consistently locating wire into the weld pool. Therefore, wire feed applications must be robust for consistency in manufacturing settings. Both pulsed- and continuous-wave lasers can be used for keyhole welding.

Other laser welding variations also capitalize on the characteristics of deep penetration laser welding. The most common combination includes hybrid laser-arc welding (HLAW). Applications of hybrid laser arc welding using the gas metal arc welding (GMAW) process include construction equipment fabrication, passenger rail cars, and shipbuilding. An example of the difference between a gas metal arc (GMA) weld, keyhole laser weld, and hybrid laser GMA weld using the same arc parameters is shown in Figure 2.3 (Martukanitz, 2005). This process improves susceptibility to poor joint fit-up by using the GMAW process with the added benefit of the laser's deep penetration capability. The hybrid characteristic allows for joining thick section components using a single weld pass and reducing total joining time. In addition, the lower overall heat input, because of a single weld pass, results in less component distortion. Other forms of laser arc hybrid welding have been studied, such as laser with GTAW (Yan et al., 2010), but HLAW–GMA is the most common and currently an industry-practiced welding method.

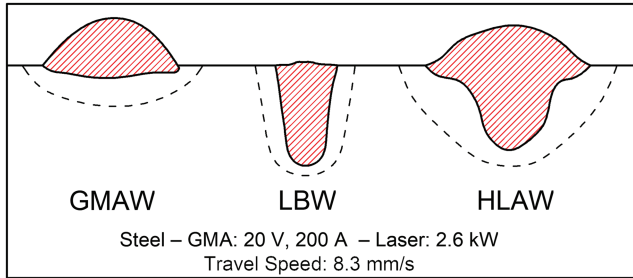


Figure 2.3 Schematic of transverse weld profiles using gas metal arc weld (left), keyhole laser weld (center), hybrid laser – GMA weld (right) based on results using 200 A, 20 V, 2.6 kW of laser power, and a travel speed of 0.5 m/min. Weld profiles from Martukanitz (2005).

Defects

As with conduction mode welds, keyhole mode welds are also prone to solidification cracking. Keyhole mode welds generate porosity defects caused by the vaporization, intrusion of atmospheric air or shielding gas, or collapse of the keyhole. The same variables resulting in solidification cracking for conduction mode welding apply to the keyhole mode. Porosity becomes more prevalent in keyhole welding and certain alloy systems are more prone to pore formation, such as aluminum (Matsunawa, 2002). Welding in partial penetration exacerbates porosity compared to full penetration welds because the vapor cannot escape through the back of the plate.

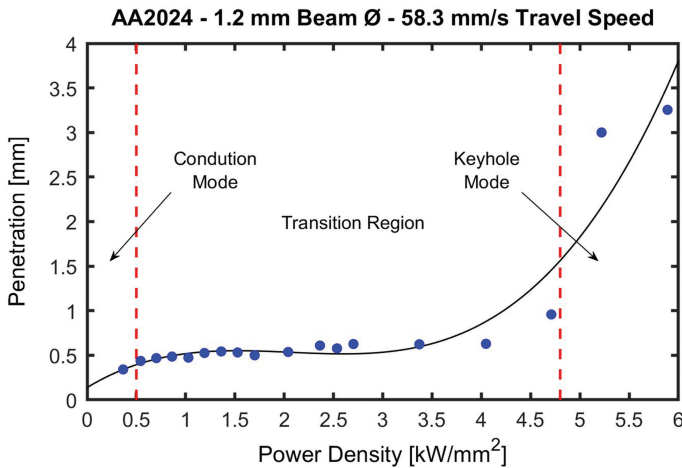
Welding at excessive travel speeds (>75 mm/s) and/or using small beam diameters (<200 μm) can also result in weld bead humping, where discontinuous globs of liquid solidify with minimal wetting (Albright & Chiang, 1988; Seiji Katayama et al., 2011; Kawahito et al., 2007; Luo et al., 2015). It is suspected that humping is largely a result of surface tension and Marangoni flows and is influenced by material composition. Kawahito et al. showed that low pressure and vacuum LBW exacerbates humping (Kawahito et al., 2007), but Katayama et al. (2011) and Börner et al. (2011) were able to eliminate humping by positioning the focal point below the plate surface, that is, under-focused.

Transition Region

It is convenient to categorize welds in one of two regimes, but the physical interactions are much more complex than simply going from minimal vaporization to boiling, which results in backpressure sufficient to depress

the weld pool. Therefore, there may be a range of parameters where a transition from conduction to keyhole welding occurs. This transition is material dependent, and the power and interaction time are such that a conduction-like mode exits** with an onset of vaporization. Figure 2.4 shows laser weld penetration for a 2024 aluminum alloy (Eurico Assuncao & Williams, 2014). For high reflectivity and high thermal conductivity alloys such as aluminum, the transition region is generally wider. Assunção et al. showed the results for aluminum and 304 stainless steel, where the power density change across the transition regime is much less for 304 stainless steel. Other factors influencing this transition include the shielding gas, plume suppression, and atmospheric pressure. Welding in reduced pressure environments accelerates vaporization and results in quicker conduction-to-keyhole mode transitions, as commonly encountered during EBW (Hochanadel et al., 2021).

Generally, the transition region is not an intentional welding regime. However, it has benefits for increasing penetration from conduction mode welding and minimizing propensity to defects formed during keyhole mode welding. There are no specific applications that rely on welding in a transition region, but any applications welding in transition-like mode are likely under the assumption that welding is within the conduction mode regime. This is mostly due to the historical classification of conduction/keyhole based on depth-to-width aspect ratios (Cunningham et al., 2019; Patterson et al., 2022).



Laser Welding Variables

The three fundamental continuous-wave laser welding variables are power, travel speed, and beam diameter. Some laser optics have capabilities to oscillate the beam through mirrors attached to galvanometers. These optics increase the variables to include oscillation pattern, oscillation diameter, and frequency. Pulsed LBW introduces additional variables that include peak power, average power, pulse shape, pulse duration, and pulse frequency.

Many equations are available in the open literature to characterize the behavior of laser welding parameters. Often, these parameters are used to gauge the weld response, mainly concerning penetration. As described previously, LBW has a lower heat input as compared to arc welding processes. Heat input is a measure of the linear energy deposited into the workpiece based on the average power and travel speed, as shown in Equation 2.1. The heat input calculation neglects beam diameter variations, intensity distributions, and energy loss. Therefore, it is a theoretical maximum sum of the energy input into the material. The heat input calculation has no term for beam diameter, and therefore, does not represent the ability to develop a weld pool.

$$HI = \frac{P_{avg}}{v_b} [J/mm] \quad \text{Equation 2.1}$$

where HI = heat input, P_{avg} = average power, and v_b = beam velocity.

Beam Intensity

The calculation for power density is shown in Equation 2.2 and is a computation of the output beam power divided by the area of the laser beam. Linearly increasing power with a constant beam diameter linearly scales the power density. Increasing the beam diameter results in an inverse square law relationship for decreasing power density. The simplistic calculation becomes more ambiguous with beams using a non-uniform beam distribution or transverse electromagnetic mode (TEM).

$$P_d = \frac{P_{avg}}{A_b} = \frac{P_{avg}}{\pi d_b^2 / 4} [W/m^2] \quad \text{Equation 2.2}$$

where P_d = power density, P_{avg} = average power, A_b = beam area, d_b = beam diameter.

The laser supply, transport fiber, and focusing optics are used to manipulate beam distributions and focal diameters. The laser optical cavity generates TEM_{xy} or standing wave patterns. Assuming the lasers with near 1 μm wavelengths, or less, delivery fibers with diameters small enough in relation to the photon wavelength will only support and transport a single TEM. This single mode results in a Gaussian intensity distribution. Large-diameter delivery fibers allow for propagating multiple TEMs through the fiber core – allowing for manipulation of intensity distributions. Manipulating the laser generation, transmission fibers, and optics allows for changing the intensity distributions to optimize weld performance. Due to absorption of 10.6 μm wavelengths, CO₂ lasers cannot be transported through silica fiber optics and TEMs are developed in the optical cavity.

Two common beam profiles used for welding are a Gaussian distribution or multi-mode, also known as “top-hat.” For multi-mode beam distributions, the uniform intensity characteristic is only maintained at or near the focal position. Figure 2.5 shows a near-Gaussian distribution and a multi-mode distribution. This distribution is the result of a transport or feed fiber that can propagate multiple TEMs. When the beam distribution is theoretically circular, the diameter and area can simply be estimated from the circular cross section. For most LBW systems, the beam diameter at the focal position is a fixed value unless the feed fiber or focusing optics are physically modified or changed. Because of the need to modify components, it is easiest to weld in a defocus condition relative to the work-piece surface. Although an easy solution, defocusing can have a drastic influence on the weld formation because the intensity distribution also changes. For the same multi-mode beam shown in Figure 2.5, the beam

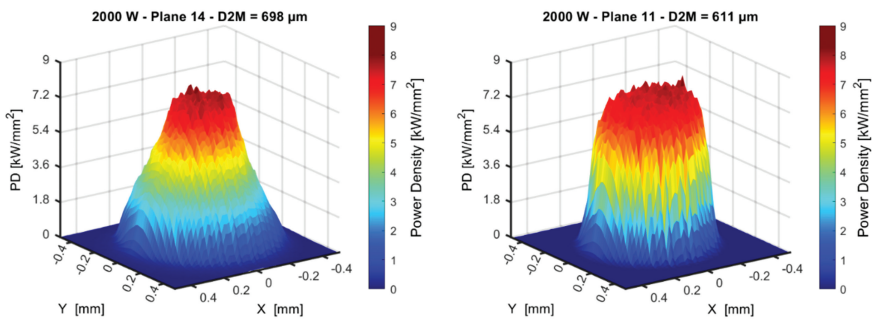


Figure 2.5 Near-Gaussian power density distribution (left). Multi-mode power density distribution near the focal position (right). Results are from a beam generated with a 600 μm feed fiber, 300 mm focal lens, and 150 mm collimator lens. The z -axis and color scale represent the power density. The D2M in each figure is the same as $d4\sigma$.

caustic measured up to 12.5 mm from each side of the focal point is shown in Figure 2.6. The color or gray scale represents the localized intensity measurements.

TEMs commonly utilized with LBW include Gaussian, multi-mode, and TEM_{01}^* . TEM_{01}^* is often referred to as “donut mode” and is a cylindrical intensity distribution. Benefits of using beam profiles that do not approach an ideal beam distribution, or Gaussian shape (TEM_{00}), are to improve welding performance. The sharp intensity peak of a Gaussian distribution can result in excessive spatter, root spiking, or keyhole instability leading to the collapse of the vapor cavity. Manufacturers now produce laser systems capable of delivering adjustable laser modes by using fibers with two cores (Feuchtenbeiner et al., 2019). An example of beam distributions using a dual-core fiber is shown in Figure 2.7. The outer core propagates the ring-shaped distribution, while the inner core delivers a Gaussian-like distribution. The power can be adjusted for each core to achieve desired results. Having both distributions in a single beam may allow for achieving keyhole mode welding while minimizing spatter formation.

For comparing laser beams and calculating the power density, beam diameters are usually measured with devices that can detect the local intensity throughout discrete measurements along the beam cross section or via a longitudinally propagating beam. Examples of these methods for measuring

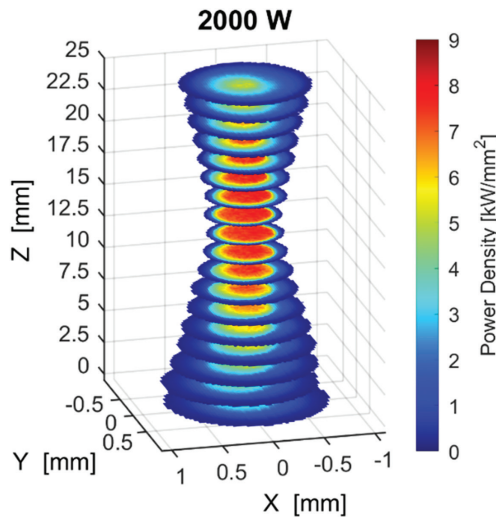


Figure 2.6 Beam caustic as measured by a rotating pinhole device (left) throughout the focal position. The color scale or gray scale represents the local intensity or power density distribution.

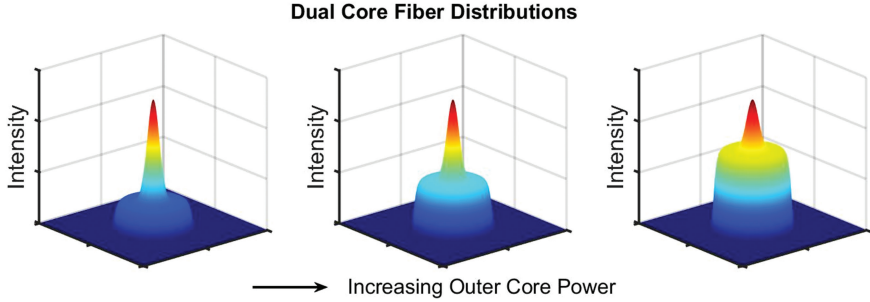


Figure 2.7 Beam distributions from a dual-core fiber with power increasing in the outer fiber from left to right.

high-power beams include a rotating pinhole device and a noncontact beam profiler using a camera to detect light scattered by air molecules. Once beam measurements are captured, there are multiple methods of measuring the beam diameter for a given plane normal to the propagation axis. Common measurements include the diameter at FWHM, the diameter where the intensity distribution decreases by $1/e^2$ or 0.186 (d_{e-2}) and the second-moment beam diameter ($d_{4\sigma}$). For an ideal Gaussian beam, d_{e-2} and $d_{4\sigma}$ will compute identical beam diameters. However, when the distributions deviate from Gaussian, each method is influenced by the profile (Pierce & Burgardt, 2014). The International Standards Organization (ISO) uses the $d_{4\sigma}$ calculation (Lasers and Laser-Related Equipment – Test Methods for Laser Beam Widths, Divergence Angles and Beam Propagation Ratios, 2005). The ISO calculations for measured beam diameters are shown in Equation 2.3–Equation 2.6.

$$\text{First moment: } X = \frac{\sum E(x,y)x}{\sum E(x,y)}, Y = \frac{\sum E(x,y)y}{\sum E(x,y)} \quad \text{Equation 2.3}$$

$$\text{Second moment: } \sigma_x^2 = \frac{\sum E(x,y)(x-X)^2}{\sum E(x,y)}, \quad \text{Equation 2.4}$$

$$\sigma_y^2 = \frac{\sum E(x,y)(y-Y)^2}{\sum E(x,y)}$$

$$\text{Overall standard deviation: } \sigma^2 = \sigma_x^2 + \sigma_y^2 \quad \text{Equation 2.5}$$

$$\text{Overall diameter: } d_{4\sigma} = 2\sqrt{2}\sigma \quad \text{Equation 2.6}$$

where X and Y = first-moment calculations in the x and y directions, σ_x^2 and σ_y^2 are the second moment in the x and y directions, σ^2 is the standard deviation, and $d_{4\sigma}$ is the second-moment beam diameter

The $d_{4\sigma}$ calculation is an integral of the measured intensity weighted based on the length from the centroid squared. For non-ideal beam shapes, each method will influence the results. Therefore, it is not possible to definitively define beam diameters, and a single computational technique must be used for direct comparison between one LBW process and TEM to another. For EBW, similar to LBW, Burgardt and Pierce (2014) demonstrated the difference in results for a non-ideal EB caustic comparison of $d_{e^{-2}}$ and $d_{4\sigma}$. Because the $d_{e^{-2}}$ calculation includes the diameter of the beam encompassing 86% of the power around the beam centroid, it is less susceptible to deviations in distributions.

Other laser beam indicators are used to determine the beam quality, which includes the divergence angle, beam parameter produce (BPP), Rayleigh length, M^2 value, or K factor. Each of these measurements can only be experimentally measured by capturing a beam caustic as shown in Figure 2.6. The divergence angle is commonly represented as a half angle and is a far-field measurement of how quickly the beam increases in size from the focal spot size. The BPP calculation is shown in Equation 2.7. A schematic indicating a focused beam and divergence half angle is shown in Figure 2.8. Another measure of how quickly the beam diverges is based on the Rayleigh length; the beam area is double the area at the focal point. The theoretical calculation for the Rayleigh length is based on the minimum beam radius and photon wavelength as presented in Equation 2.8, and the calculation for the beam radius at the Rayleigh length is shown in Equation 2.9. Ideally, the beams used for welding should have long Rayleigh lengths. Short Rayleigh lengths, because of a large diverge angle, are prone to inconsistent welds caused by slight variations in material location relative to the focal position. A small beam diameter and/

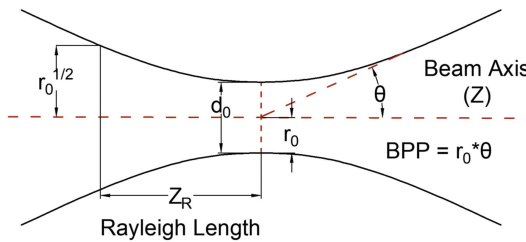


Figure 2.8 Schematic of a focused beam propagating through the longitudinal (z -axis). Locations are indicated for the focused beam diameter (d_0), divergence half angle (θ), and Rayleigh length (Z_R).

or diverge angle also contribute to a smaller BPP. The lower the BPP value, the higher the quality beam due to a smaller spot size and/or reduced divergence.

$$BPP = r * \theta [mm * mrad] \quad \text{Equation 2.7}$$

where r = beam radius at the focal point and θ = divergence half angle.

$$Z_R = \frac{\pi r^2}{\lambda} [mm] \quad \text{Equation 2.8}$$

where r = beam radius at the focal point, π = pi, and λ = wavelength.

$$r_R = \sqrt{2} * r_0 [mm] \quad \text{Equation 2.9}$$

where r_R = beam radius at the Rayleigh length and r_0 = minimum beam radius at the focal point.

The M^2 value is based on the ratios of BPP of an actual beam to that of the ideal Gaussian beam as shown in Equation 2.10. Gaussian beams with a high beam quality have M^2 values that approach one, the theoretical maximum. Beams with multiple modes, such as the multi-mode distribution shown in Figure 2.5, result in M^2 values of approximately 20–30 and are far from a theoretically ideal beam. However, these “non-ideal” distributions may improve welding performance. Another variation in the M^2 value is the beam propagation factor, K . The K -value is the inverse of the M^2 value as shown in Equation 2.11.

$$M^2 = \frac{\pi r_0 \theta}{\lambda} [\text{unitless}] \quad \text{Equation 2.10}$$

where M^2 = M -squared value, r_0 = minimum beam radius at the focal point, π = pi, and λ = wavelength.

$$K = 1/M^2 [\text{unitless}]. \quad \text{Equation 2.11}$$

where K = beam propagation factor and M^2 = M -squared value.

Although these beam characteristics define the properties of the beam, the power density is only based on a computed beam diameter and does not incorporate localized intensity distributions. For simple power density analysis, it is necessary to assume some value of uniformly distributed intensity, but caution should be taken prior to directly comparing weld results from different laser processes.

Efficiency

The simplest form of computing the LBW efficiency is shown in Equation 2.12 and computed based on the power absorbed by the workpiece and the total power applied to the workpiece (Fuerschbach, 1996; Swift-Hook & Gick, 1973). However, this requires the arduous task of measuring the power absorbed by the workpiece. The complexity of the LBW process involves interactions with solid, liquid, gas, and sometimes plasma phases. In a steady-state welding condition, there will be some fraction of beam reflection, scattering, and/or absorption in the vapor plume. In addition to radiation losses, these all reduce the energy transferred to the workpiece and result in reduced energy transfer efficiency. This efficiency, commonly known as beam coupling, is influenced by process parameters such as power density, photon wavelength, angle of incident radiation, material chemistry, and surface roughness.

$$\eta = \frac{P_i}{P} [\text{unitless}] \quad \text{Equation 2.12}$$

where η = energy transfer efficiency, P_i = power absorbed, and P = total power applied.

The energy transfer efficiency accounts for the energy lost due to interactions of the laser with the material and atmosphere but is not a measurement of the energy needed to generate a weld pool. Thermal conduction of the energy into the workpiece results in significant energy losses during welding, and overcoming latent heat during phase changes requires additional energy input. Ream reported an equation to represent the melting efficiency based on the transverse cross-sectional area of the weld, weld speed, and power, as shown in Equation 2.13 (Ream, 2004). As computed, this is not a true efficiency with resultant units of millimeters cubed per joule. In addition, only accounting for the cross section does not account for the length or three-dimensional (3D) molten volume.

$$\eta_m = \frac{A_w v_b}{P} [mm^3/J] \quad \text{Equation 2.13}$$

where η_m = melting efficiency, A_w = transverse cross-sectional area of weld, v_b = weld velocity, and P = total power input.

If the molten volume can be determined, it is possible to estimate the melting efficiency by computing the theoretical amount of energy required to melt a given volume. The equation to compute melting energy for continuous-wave laser welding is shown in Equation 2.14. Although with limited success, weld

pool decanting was shown to approximate weld pool volumes for conduction mode welds (Tate Patterson, 2021).

$$Q_m = Av_b \rho \left(\int_{T_0}^{T_m} C_p dT + \Delta H_m \right) [J] \quad \text{Equation 2.14}$$

where Q_m = energy to melt, A = transverse weld area, v_b = weld velocity, T_m = melting temperature, T_0 = plate temperature, C_p = specific heat capacity, T = temperature, and H_m = latent heat of fusion.

Manipulation of Variables

To better quantify laser weld behavior, analysis attempts have been performed using equations relating to beam diameter, travel speed, and/or power. One of those variables is the interaction time or the time that a finite point is exposed to the beam energy (Assunção, 2012; Courtois et al., 2013; Hann et al., 2011; Suder & Williams, 2012). The interaction time calculation is shown in Equation 2.15. It should be noted that this is only for continuous-wave LBW and not pulsed LBW. For circular-shaped beams, this incorrectly assumes that the interaction time is the same across the beam area. The actual interaction time is relative to the position of the beam along the axis perpendicular to the direction of travel. Figure 2.9 shows a schematic of how the actual interaction time would change at different locations across the beam. Literature implementing interaction time analysis must assume that the error in the varying length across the spot size is negligible. Suder and Williams reported that continuous-wave interaction time may be comparable to a pulse period during pulsed LBW (Suder & Williams, 2012).

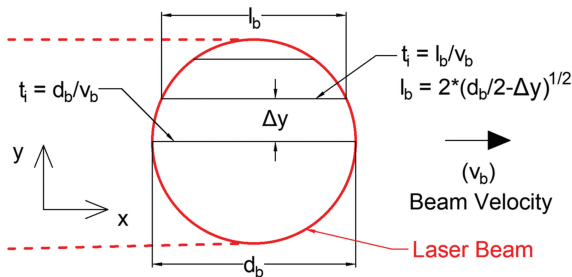


Figure 2.9 Schematic showing the change in interaction time at different beam locations for a circular beam traveling left to right.

$$t_i = d_b / v_b \text{ [s]} \quad \text{Equation 2.15}$$

where t_i = interaction time, v_b = beam velocity, and d_b = beam diameter.

Building on the assumption of interaction time, the specific point energy (SPE) is the interaction time multiplied by the beam power as shown in Equation 2.16. Conceptually, this is the total energy applied to a point along the beam center. To compute an energy density, the SPE can be divided by the beam area, or the power density can be multiplied by the interaction time as shown in Equation 2.17. Suder and Williams described relationships between constant interaction times and power density relating to increasing weld penetration. They also reported a need to achieve a threshold energy density and interaction time to initiate keyhole mode welding (Suder & Williams, 2012).

$$\text{SPE} = \frac{P d_b}{v_b} = P t_i \text{ [J]} \quad \text{Equation 2.16}$$

where SPE = specific point energy, P = power, d_d = beam diameter, v_b = beam velocity, and t_i = interaction time.

$$\text{ED} = \frac{P d_b}{A_b v_b} = \frac{\text{SPE}}{A_b} = P_d t_i \left[\frac{\text{J}}{\text{m}^2} \right] \quad \text{Equation 2.17}$$

where ED = energy density, P = power, d_d = beam diameter, v_b = beam velocity, A_b = beam area, SPE = specific point energy, t_i = interaction time, and P_d = power density.

Parameter Influences

The following sections report data from literature relating to how process parameters influence weld formation. It is difficult to directly compare many LBW results based on the various lasers, spot sizes, focal conditions, plume suppression, and many other configurations used during investigations. However, the trends are generally comparable between similar process setups.

Power

Although literature may report simplistic trends such as linearly increasing penetration with increased power, these trends seldom exist over a wide power

range. Penetration is largely influenced by the beam diameter, power density, plume suppression techniques, and shielding gases for a given material. Some reports have suggested that weld penetration increases proportionally to power raised to two-thirds ($d_p \propto P^{2/3}$) at low welding speeds (<100 mm/s), and penetration is linearly proportional to power at high welding speeds (>100 mm/s).

Figure 2.10 shows Ti-6Al-4V results for laser beam penetration and cross-sectional area for numerous travel speeds and increasing power with a constant beam diameter. These welds were deposited in the atmosphere with argon shielding (Tate Patterson, 2021). The penetration profiles indicate a distinct nonlinear increase of penetration depth with increased power for all travel speeds analyzed. At travel speeds below 100 mm/s, the trends were consistent with the $d_p \propto P^{2/3}$ relationship. However, the 100 mm/s does not show a linear penetration increase. However, the measurements of the transverse cross-sectional area show a nearly linear increase in area with increased power. Although convenient, none of these approximations consider the complexity of the process relative to changes in vaporization, beam coupling efficiency, and thermal conductive losses.

As previously shown in Figure 2.4, welding using lower power density or high travel speeds can influence the penetration behavior and result in wide transitions between conduction and keyhole mode welding without clear indication of a transition (Assunção, 2012). The complex processes of vaporization, beam attenuation, and surface tension-driven fluid flows are all based

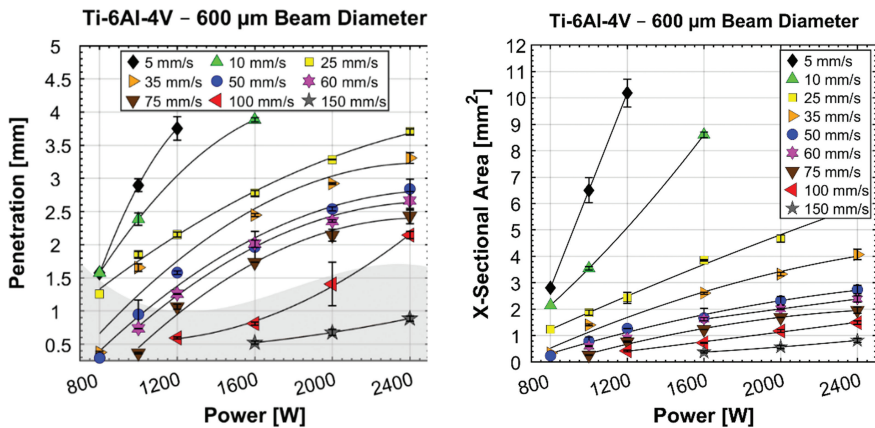


Figure 2.10 Ti-6Al-4V weld penetration versus power for different welding speeds (left). The shaded region approximates conduction mode welds. Ti-6Al-4V fusion zone transverse cross-sectional area versus power for different travel speeds (right) (Tate Patterson, 2021).

on material properties, impurity contents, and process parameters (Kou et al., 2011; Limmaneevichitr & Kou, 2000).

Beam Diameter

As with varying power, changing the beam diameter influences the power density applied to the workpiece. Beam diameters used for LBW commonly range from 300 to 1500 μm in diameter. Beam sizes below 300 μm are typically used for laser beam cutting due to high intensity and excessive vaporization. The high degree of vaporization allows for sufficiently penetrating through materials and generating thin melt layers, which are undesirable on cut surfaces. Because the beam diameter results in an inverse square law relationship, power density reduces accordingly with an increased beam diameter. Thus, penetration trends similarly decrease with increasing beam diameter. Usually, the focal point diameter is fixed for a welding system unless the laser has an adjustable mode, adjustable optics, or a dual-core fiber. Changing the beam delivery, optics, or adjusting the beam mode will modify more than just the spot size and will influence the distribution, BPP, and M^2 value.

Focal Position

The placement of the focal position relative to the plate surface will influence the resultant weld formation depending on the defocused position. A common technique to increase penetration is to weld in an under-focused condition, where the focal point is positioned slightly below (or under) the plate surface (Seiji Katayama et al., 2011; Webber et al., 2011). When welding in an over-focused condition, or focal point above the plate surface, the diameter projected onto the workpiece will increase – reducing the power density and resulting in reduced weld penetration. An example of a slight increase in spot weld penetration at different focus positions is shown in Figure 2.11. Likewise, sufficiently under-focusing the beam will also result in reduced weld penetration as the energy is lost prior to reaching the maximum power density.

Travel Speed

It is well documented that the trend for increasing travel speed will result in an exponential decrease in weld penetration. Numerous LBW papers have been published in open literature indicating this relationship (Assunção, 2012; Baardsen et al., 1973; Chiang & Albright, 1993; Cho et al., 2009; Locke & Hella, 1974; Quintino et al., 2007; Suder & Williams, 2014). Higher travel

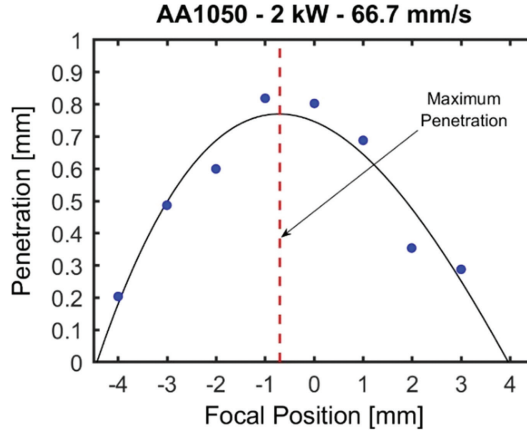


Figure 2.11 Influence of the focal position on weld penetration in aluminum alloy 1105 using a beam with 6 mm*mrad BPP and 2.8 mm Rayleigh length at 2 kW and a travel speed of 66.7 mm/s. Data extracted from T. Sun et al. (2021).

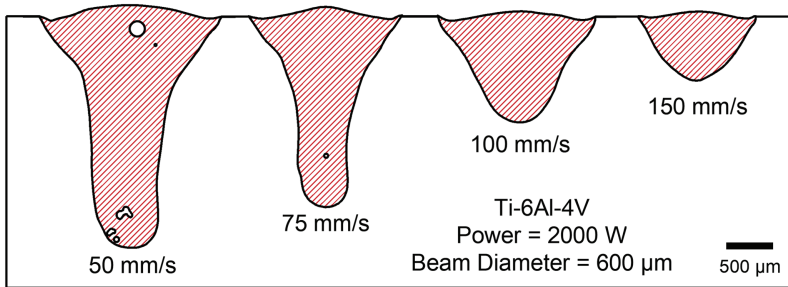


Figure 2.12 Ti-6Al-4V weld geometry versus travel speed using a multi-mode beam profile at sharp focus. Data extracted from Tate Patterson (2021).

speeds result in less conductive heat loss by reducing the time but require much higher power to maintain weld penetration. One influence of high travel speeds is to effectively “out-run” the vapor plume, which may result in less beam attenuation (Börner et al., 2013; Jiang et al., 2019; Seiji Katayama et al., 2001; Nakamura et al., 2015; Reisingen et al., 2016). An example of travel speed on the shape of the transverse weld section is shown in Figure 2.12. These welds were performed on Ti-6Al-4V within the atmosphere using argon shielding (Tate Patterson, 2021). It is likely that the 50–100 mm/s welds cross through a transition region, with both keyhole and conduction-like features present in the 100 mm/s weld.

There is potential that travel speeds can become too fast, or interaction times too short, and Baardsen et al. reported that increasing travel speed

resulted in less energy effectively “captured” within the keyhole (Baardsen et al., 1973). Without sufficient time for beam interaction with the workpiece, insufficient heating ensues and a fraction of the beam reflects off the solid surface. Gladush and Smurov reported the results of the melting efficiency relative to weld speed and showed that too slow or too fast of travel speeds reduced the efficiency (Gladush & Smurov, 2011).

Beam Manipulation

Using galvanometers to manipulate the laser beam adds complexity to the weld process but may prove advantageous for welding certain alloys or alloy combinations. Reports have indicated that beam oscillation can improve the weld surface appearance, reduce spatter, and result in equiaxed grain formation in the fusion zone. An example of the influence of beam oscillation for autogenous LBW on Inconel® 740H® is shown in Figure 2.13. The transverse cross-sectional results indicate how drastic beam oscillation can change the weld profile and lead to issues such as excessive undercut. Wang et al. showed that circular beam oscillation, compared to a stationary beam, generated more equiaxed grains and the smoothest weld surfaces in 6061-T6 aluminum (Wang et al., 2016). An example showing porosity improvement because of beam oscillation in 6061 is shown in Figure 2.14. Although the longitudinal section

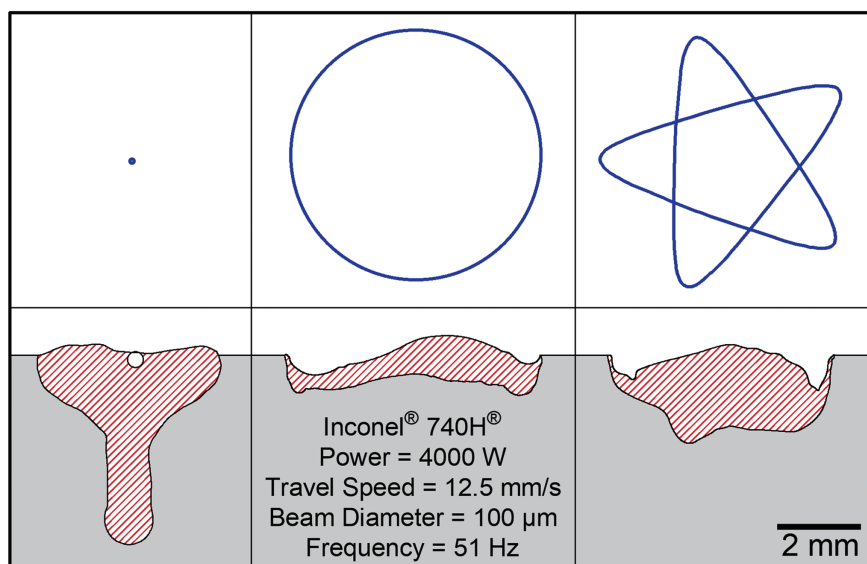


Figure 2.13 Influence of laser oscillation patterns at 6 mm amplitude and 51 Hz frequency.

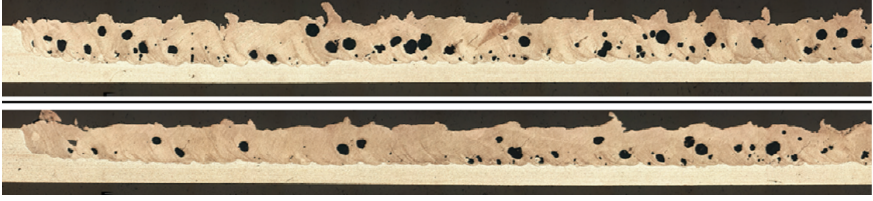


Figure 2.14 Example of porosity improvement with a 100- μm -diameter beam using no beam oscillation (top) and circular beam oscillation at a frequency of 1 kHz and amplitude of 50 μm (bottom). Welds were made up of aluminum alloy 6061 (Thiel et al., 2012).

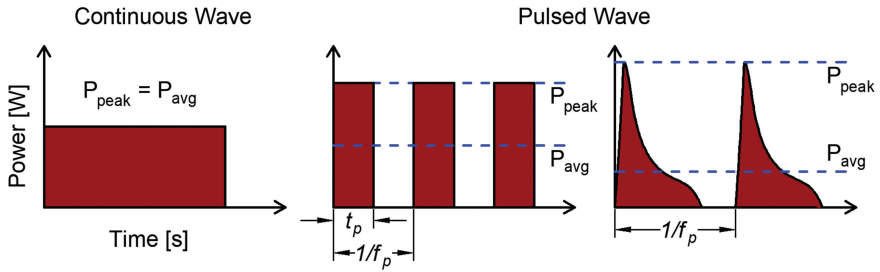


Figure 2.15 Schematic representation of continuous-wave versus pulsed-wave outputs.

remained full of porosity, there was a noticeable decrease in the total quantity of pores.

Pulsed Laser Beam Welding

The additional variables associated with pulsed LBW include peak power (P_{peak}), average power (P_{avg}), pulse duration (t_p), and pulse frequency (f_p). A schematic representation of continuous-wave and pulsed-wave power outputs is shown in Figure 2.15. Depending on the laser, different pulsed waveform configurations are possible for customizing the ramp rates and power throughout the pulse. Assuming square wave pulses, additional variables that defined pulsed laser welding including the pulse energy and average power are shown in Equation 2.19 and Equation 2.18, respectively. For complex waveforms, the equations become integrals to compute the area (energy) under the power–time curves. Because successive pulses are used during the weld travel, it is necessary to adjust pulse durations and frequencies such that sufficient overlap

develops a structurally sound and/or hermetically sealed joint. The equation to predict pulse overlap is shown in Equation 2.20.

$$E_p = P_{peak} t_p [J] \quad \text{Equation 2.18}$$

where E_p = pulse energy, P_{avg} = average power, and t_p = pulse time.

$$P_{avg} = E_p f_p [W] \quad \text{Equation 2.19}$$

where P_{avg} = average power, E_p = pulse energy, and f_p = pulse frequency.

$$\% \text{Overlap} = \left(1 - \frac{v_b / f_p}{d_b + v_b t} \right) * 100\% [\%] \quad \text{Equation 2.20}$$

where v_b = weld velocity, d_b = beam diameter, and f_p = pulse frequency.

Pulsed LBW can be advantageous compared to continuous-wave LBW by improving or having greater flexibility to adjust the coupling efficiency, reduce overall heat input, improve weld surface appearance, or change penetration depth. Considering a single spot weld, the energy required to melt the fusion zone volume can be approximated via Equation 2.21. This theoretical energy calculation can be used with the actual energy to determine the efficiency of the process based on Equation 2.12. Equations are also available for determining the energy to melt based on traveling, pulsed laser welds as reported by Sabbaghzadeh et al. (2008).

$$Q_m = V_w \rho \left(\int_{T_0}^{T_m} C_p dT + \Delta H_m \right) / E_p [J] \quad \text{Equation 2.21}$$

where Q_m = energy to melt, V_w = weld volume, v_b = weld velocity, T_m = melting temperature, T_0 = plate temperature, C_p = specific heat capacity, T = temperature, H_m = latent heat of fusion, and E_p = pulse energy.

Using the pulsed laser welding variables, Corrado et al. showed additional analysis through a developed “pulse power factor model.” They reported that the peak pulse power controls weld penetration, whereas interaction time controls the weld width (Corrado et al., 2017). A comparison of pulsed laser and continuous-wave laser weld penetration using nearly the same beam diameter and travel speed is shown in Figure 2.16. The penetration improvement by using a pulsed laser is evident and is further improved at higher power densities. For pulsed LBW, the results in Figure 2.16 also show no or negligible

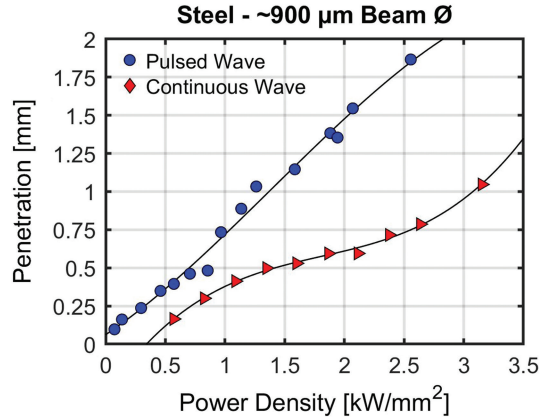


Figure 2.16 Comparison of pulsed LBW penetration and continuous-wave LBW penetration. Beam diameters were 0.9 mm for the pulsed laser and 0.95 mm for the continuous-wave laser. Travel speeds were 45 mm/s for the pulsed laser and 47.5 mm/s for the continuous-wave laser. Data extracted from Eurico Assuncao and Williams (2013).

transition region between conduction and keyhole mode welding (Eurico Assuncao & Williams, 2013). Other reports have shown that increasing average power for pulsed LBW created a nearly linear increase in the fusion zone cross-sectional area (Kumar & Sinha, 2019).

Pulsed LBW can be susceptible to solidification cracking defects caused by rapid solidification rates (Lippold, 1994), but the ability to alter pulse waveforms can help control the solidification rate, cooling rate, and pore formation. Sun et al. showed the ability to change the microstructure and weld metal hardness by changing the pulse frequency when welding a dual-phase steel (Sun et al., 2016), and Chludzinski et al. reported minimizing cracking and pore formation through pulse parameter adjustments when welding an aluminum alloy (Chludzinski et al., 2022).

Shielding Gas

Applying the proper inert or nonreactive shielding gas improves weld quality when laser welding any metallic alloy. Although shielding gas application is not always a common industrial practice because it adds production costs, e.g., the automotive industry regularly performs unshielded laser welding. However, alloys highly susceptible to oxidation, such as aluminum and

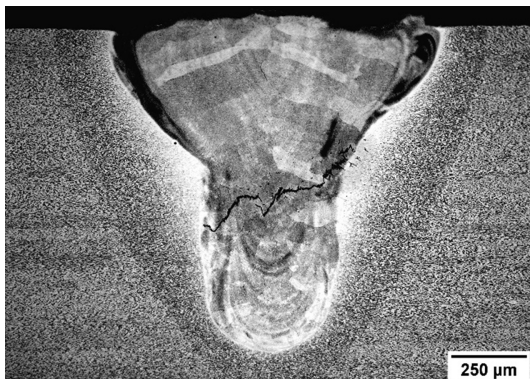


Figure 2.17 Oxidized Ti-6Al-4V laser weld showing transgranular cracking caused by atmospheric contamination.

titanium, should be carefully shielded when welded in atmosphere. Not using or using improper shielding gas can result in brittle oxides/nitrides, such as alpha case in titanium alloys. An example of a poorly shielded Ti-6Al-4V weld resulting in brittle, solid-state cracking shortly after welding is shown in Figure 2.17.

Shielding gas has a broader influence than solely protecting the weld against oxidation. Gases can affect the energy transferred to the workpiece, weld metal porosity, and solid-state microstructure formation. The flow rates and angle of flow can also influence the shielding quality, plume formation, and weld penetration as shown by Sibillano et al. (2006). Helium was shown to increase weld penetration compared to argon and CO₂ shielding gases for a 10.6 μm laser, with argon having the lowest penetration (Chiang & Albright, 1993).

Elmer et al. reported that A36 steel and 304L stainless steel showed minimal or no porosity when laser welded with nitrogen shielding gas but exhibited high levels of porosity when welded using argon shielding. Pure nickel was prone to oxidation regardless of argon or nitrogen shielding, but 21-6-9 stainless steel showed little porosity with either argon or nitrogen (Elmer et al., 2015). Duplex stainless steels are prone to nitrogen loss when welding with non-nitrogen-containing shielding gases (Keskitalo et al., 2015). Because nitrogen is a potent austenite stabilizing element, austenite phase fractions decrease within the fusion zone, leading to deleterious corrosion properties. The use of pure nitrogen shielding for duplex stainless steel was shown to increase austenite fractions versus pure argon shielding (Baghdadchi et al., n.d.; Keskitalo et al., 2015).

In addition to shielding gas, atmospheric laser welding is often applied with plume suppression techniques that incorporate high pressure, high velocity gas jets to expel particles out of the beam path. Removing the plume from the beam will influence weld penetration by eliminating beam attenuation. It should be noted that using a “cross-jet” or “air-knife” is not for oxidation protection but to improve weld penetration by eliminating beam interactions and improving process consistency. This is further discussed in Section 5.3.

Physical Interactions

The following sections review the physics associated with laser–material interactions. LBW is a complex process that is still not fully understood – mostly because of the inability to experimentally measure the process characteristics *in situ*, such as photon interactions, weld pool development, molten material properties, and plume/plasma formations. Specifically, it includes reflectivity, vaporization, plume formation, and weld pool fluid flows.

Reflectivity

Reflectivity is wavelength dependent and increases for longer wavelengths (American Welding Society, 2007; Steen & Mazumder, 2010; Xie et al., 1997). An example of absorption, the opposite of reflectivity, for molten iron is shown in Figure 2.17 (Rodrigues et al., 2018). The Fresnel absorption is the light reflected/transmitted across an interface of two materials, such as air and metal. In Figure 2.17, the plot of Fresnel absorption for p-plane polarized light shows the lowest reflectivity at an angle of incidence near 80° for a $1\text{ }\mu\text{m}$ wavelength. These calculations are based on assumptions for the refractive indexes. Commonly, LBW will be performed with the optics at an angle between 80° and 90° relative to the plate surface. This can help with energy absorption and is generally performed to prevent photons from reflecting back through the optics into the lasing cavity.

Reflectivity is particularly an issue with metallic alloys and related to the alloy’s conductivity and surface finish (Hipp et al., 2019). An alloy with a higher thermal and electrical conductivity will have higher reflectivity, e.g., aluminum and copper, and a comparison of pure alloy absorptivities at room temperature for a $1\text{ }\mu\text{m}$ wavelength is shown in Figure 2.18. The photon wavelengths commonly used for welding range from 0.4 to $10.6\text{ }\mu\text{m}$. The long wavelength CO_2 lasers were prone to high amounts of reflection and minimal energy absorption into the workpiece. With the advent of higher power

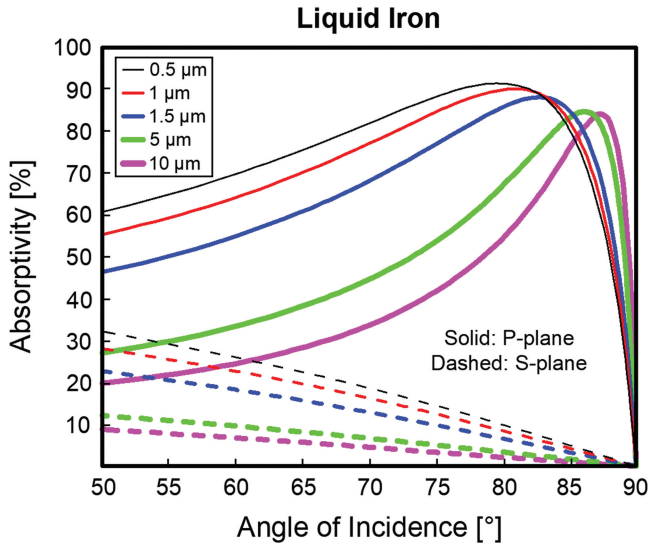


Figure 2.18 Fresnel absorption at varying wavelengths and incident angles for liquid iron. The solid lines correspond to the p-plane polarization, and the dashed lines correspond to the s-plane polarization.

YAG, fiber, and diode lasers, wavelengths as low as 450 nm are commercially available for efficiently welding on highly reflective alloys. The opposite of reflectivity is absorptivity. The absorptivity is closely proportional to the square root of electrical resistivity (Elmer et al., 2011; Mazumder, 1982; Webber et al., 2011). As with electrical or thermal conductivity, the reflectivity decreases as the material temperature rises. The photons are believed to only interact with the atoms and electrons at depths near 10–100 nm below the metal surface (Elmer et al., 2011; Mazumder, 1982; Webber et al., 2011). The shorter the wavelength, the deeper the depth of interaction and energy absorption into the workpiece.

Methods to Reduce Beam Reflection

As previously discussed, beam reflection can be minimized by reducing the wavelength. This is one of the benefits that diode, disk, fiber, and Nd:YAG lasers (near 1 μm) have over CO₂ lasers (10.6 μm). However, this is not practical for a given laser system. If deep penetration welds are possible, the vapor cavity effectively “traps” the photons to improve efficiency (Bergmann et al., 2015; Hess et al., 2011). The additional absorbed energy with a keyhole is related to the many reflections off the keyhole wall. Fresnel absorption is not dependent on the laser intensity, but increasing the number of photon–wall

interactions via the vapor cavity allows for eventually absorbing the photons, resulting in greater overall energy absorption (Dowden, 2009).

Tilting the laser focusing optics can improve the Fresnel absorption for p- and s-polarization. This is likely most helpful during conduction mode welding where there is minimal depression of the weld pool surface. However, during deep penetration welding, the vapor cavity will form a front wall angle that may influence the absorption without the need to manipulate the beam angle. X-ray imaging has shown the keyhole front wall inclinations varying from nearly vertical, or 90° , to 77° with travel speeds ranging from 17 mm/s to 167 mm/s travel speed, respectively (Miyagi & Wang, 2020).

Pulsed LBW can significantly improve beam coupling performance by imparting a short-duration, high-energy pulse to effectively vaporize the material and allow for the energy to absorb deeper into the workpiece (Olsen & Alting, 1995). High repetition rates can maintain the keyhole, but insufficient pulse frequencies and durations may be prone to keyhole collapse and porosity formation.

Vaporization

Vaporization is influenced by the chemical composition, degree of superheating, fluid flows, and elemental diffusivities. Without vaporization, there is no ability for depression of the weld pool. The vapor develops a back pressure on the liquid to maintain a compressive force against the keyhole wall (Matsunawa & Semak, 1997). The recoil pressure is necessary to depress the molten weld pool and must exceed the surface tension and hydrostatic forces (Gladush & Smurov, 2011; Matsunawa & Semak, 1997). For welding parameters producing a keyhole wall angle above 80° , deep penetration and high aspect ratio welds are reported to remain stabilized, indicating a complete transition into keyhole mode welding (Gladush & Smurov, 2011). The issues with fully understanding keyhole dynamics and vaporization are related to the influence of other parameters, such as shielding gas flow rates (Arata, 1986).

Vaporization is mostly attributed to keyhole formation but has also been shown to be relevant in conduction mode and transition regime welding. The vapor pressure above the weld pool influences the pool's depression, fluid flow, and process stability (Russo et al., 1990). To further understand keyhole evolution versus process parameters, inline coherent imaging using an optical computed tomography approach was performed to map the 3D nature of the vapor cavity. The results of this study are shown in Figure 2.19 (T Patterson et al., 2022). The drastic difference of the vapor cavity between two commercial alloys is shown for the same LBW parameters. The results from Patterson et al. also

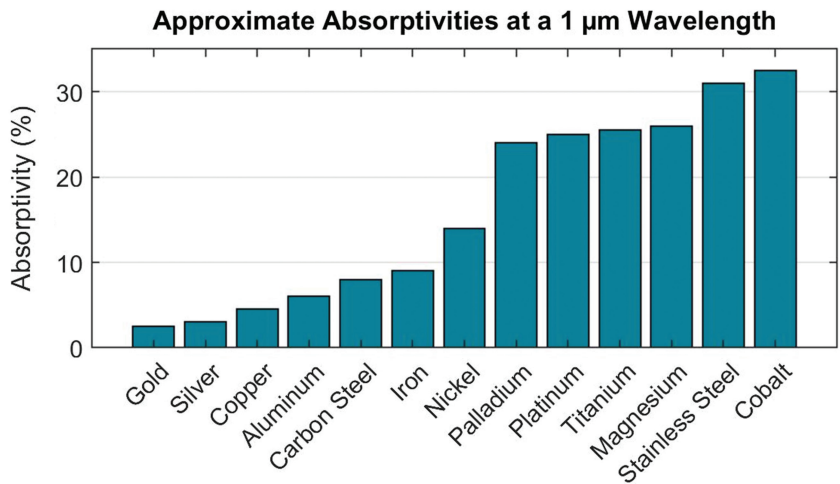


Figure 2.19 Absorptivity for different metals near 20°C at a 1 μm wavelength. Data extracted from Quazi et al. (2020).

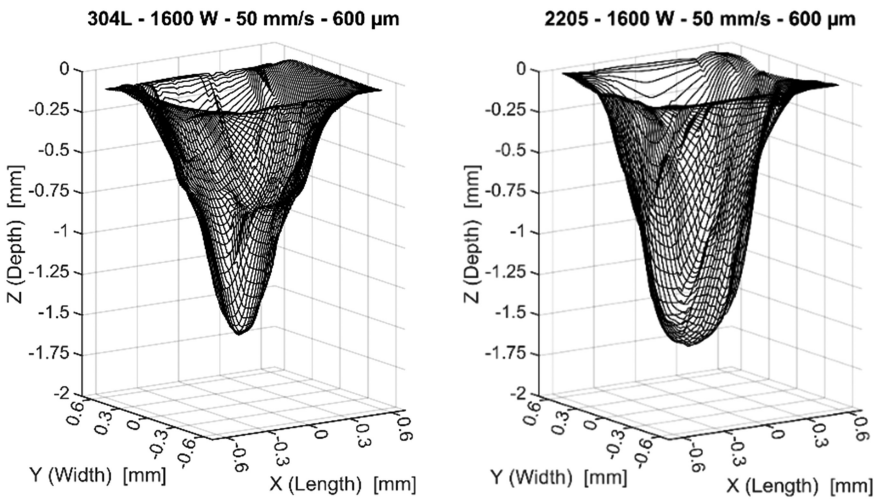


Figure 2.20 Three-dimensional representation of keyholes from optical computed tomography analysis (T Patterson et al., 2022).

showed that vapor depressions are present in conduction mode welds, which is consistent with other observations from Cunningham et al. (2019).

Additional vaporization behavior is shown in Figure 2.20 based on the location of the vapor cavity relative to the process beam position. With increased travel speed, the keyhole or vapor cavity increased the distance from

the process beam position and “lagged” behind the incident energy. As the weld transferred from a keyhole to conduction mode weld, the lag of the vapor depression maintained a distance comparable to the beam radius, which was 300 μm (Patterson et al., 2022).

Vapor Plume

The nature of vaporization during LBW can influence the energy absorption mechanisms and result in power loss through beam attenuation. The plume disrupts the beam ultimately resulting in reduced weld penetration and fusion zone area (Greses et al., 2004). The attenuation mechanisms include inverse bremsstrahlung, absorption, and/or scattering.

With 10.6 μm lasers, the vapor plume above the weld is believed to reach temperatures high enough to be a plasma. Temperature measurements have been reported as high as 17,000 K (Rockstroh & Mazumder, 1987). The molecules and atoms heated above the ionization temperature are stripped of electrons to form a state of charge particles. Free electrons within the plasma plume can absorb the photons with longer wavelengths resulting in greater absorption. This radiation absorption phenomenon is known as inverse bremsstrahlung or braking radiation and may scale with the square of the wavelength (Fabbro et al., 2016). For near 1 μm wavelengths, it is believed that the temperatures are not high enough to sufficiently generate a plasma with literature-reported measurements of approximately 3000–6000 K (Cai et al., 2018; Katayama, 2013; Shcheglov et al., 2013; Zou et al., 2016). Therefore, inverse bremsstrahlung influences are usually neglected. However, beam attenuation can still occur through other mechanisms such as Rayleigh and/or Mie scattering.

Lasers with near 1 μm wavelengths and lower are mainly influenced by Rayleigh and Mie scattering. Both scattering mechanisms are dependent on the wavelength and particle size. If the particles are much smaller (less than 1/10) than the incident photon wavelength, the result is Rayleigh (elastic) scattering. For Rayleigh scattering, the shorter the wavelength in comparison to the particle size, the more dominant the effect of scattering – inversely proportional to the fourth power of wavelength (Greses et al., 2004). The amount of beam attenuation is dependent on the particle sizes formed and the wavelength of the laser. Lost beam intensity is generally represented by the Beer–Lambert law as shown in Equation 2.22 (Scholz et al., 2015). The laser intensity reaching the workpiece is exponentially dependent on the length of beam interaction with the plume and an attenuation coefficient. The attenuation coefficient is based on the particle density, particle size, and photon wavelength and may include analysis for Rayleigh and Mie scattering (Hansen & Duley, 1994;

Scholz et al., 2015). Mie scattering occurs when the particle sizes are similar in size to the photon wavelength. Reported particle sizes have shown to range from nearly 10 nm (Scholz et al., 2015) to 50 nm (Greses et al., 2004). Based on these measurements, it can be concluded that most scattering for near 1 μm wavelengths is via Rayleigh scattering.

$$I_t(x) = I_0 e^{-\alpha x} [\text{W/mm}^2] \quad \text{Equation 2.22}$$

where I_t = transmitted intensity, I_0 = incident intensity, x = length of laser interaction with the plume, and α = attenuation coefficient.

For improving laser welding efficiency and penetration, it is advisable to minimize the beam–plume interaction. One method of minimizing the plume interaction is to remove the plume from the incident beam with a jet of compressed gas via a “cross-jet” or “air-knife.” This is a relatively simple approach to effectively blow the particles out of the beam path. An example of the influence of the air-knife height and the increase in penetration is shown in Figure 2.21 (Zou et al., 2016).

Another method to remove the plume is to perform LBW under reduced pressure or in a vacuum environment. Reduced pressure welding can significantly improve weld penetration and reduce weld porosity (Elmer et al., 2015, 2022). Figure 2.22 shows the results of weld penetration depth relative to pressure. With welding speeds less than 16 mm/s, pressures below 100 kPa increased weld penetration (Seiji Katayama et al., 2011; Reisgen et al., 2016). At pressures near and below 1 kPa, further pressure reduction showed minimal influence regarding increased penetration. This threshold has been reported as occurring at pressures as high as 10 kPa (75 torr). As the welding speed is increased, the influence of reduced pressure was negligible near a 100 mm/s travel speed.

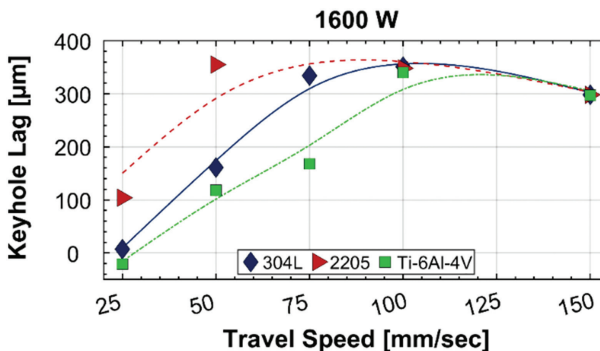


Figure 2.21 Keyhole lag versus travel speed for three commercial alloys using a 600 μm beam diameter (T Patterson et al., 2022).

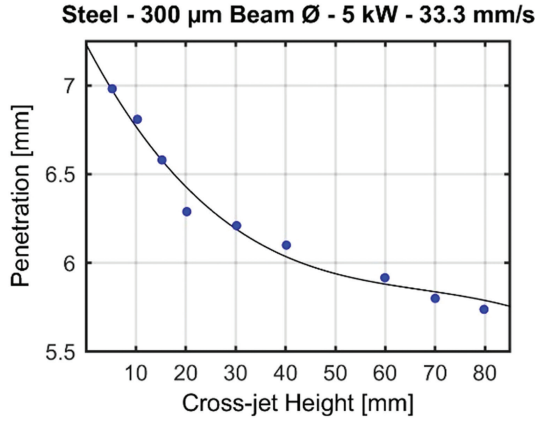


Figure 2.22 LB penetration versus cross-jet height for a fiber laser (1070 nm) using a 300 μm beam diameter. Data extracted from Zou et al. (2016).

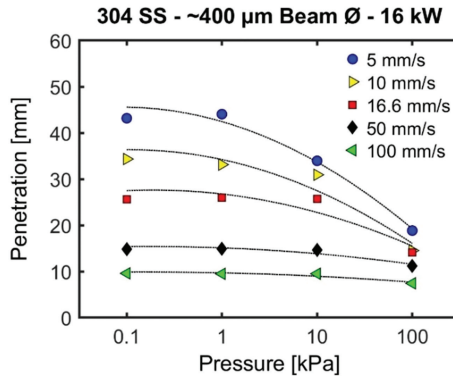


Figure 2.23 Penetration depth versus chamber pressure for a disk laser (1030 nm) at varying travel speed. Data extracted from Luo et al. (2015).

As in EBW, the vacuum environment creates a steep pressure differential and decreases the particle density in the plume. Börner et al. (2013) showed the difference in plume during reduced pressure LBW on steel, and these results are shown in Figure 2.23. A drastic reduction in the plume occurred at a pressure of 1 kPa, where only a slender region of the plume emitted visible radiation. In comparison to the plume at 100 kPa, most of the beam path is visible with high amounts of spatter.

At lower pressures, it is believed that there are fewer total particles and a lower particle density within the beam path. High travel speeds may enable the beam to “out-run” the plume, resulting in no particle interactions with

the beam. Therefore, lower pressures at higher travel speeds (>50 mm/s) likely have little influence at further eliminating particle interactions as shown in Figure 2.22. The ability for LBW to reach maximum penetration without reaching medium and high vacuum is a major benefit in comparison to EBW. Equipment costs are much higher to employ extensive vacuum systems, or manufacturing efficiency is greatly reduced to wait for long pumping times.

Fluid Flow

The fluid flow in the weld pool defines the weld shape and solidification conditions. The resultant flows are influenced by the material composition, thermal gradients, and vapor interaction. Forces acting on the liquid include, but are not limited to, Marangoni (temperature-dependent surface tension), gravity, buoyancy, and vapor pressure.

Welds where minimal amounts of vaporization are present, or conduction mode, result in shapes defined by the convective fluid flows. The composition and impurity contents influence the surface tension forces and can reverse fluid flows to result in drastically different aspect ratios (Heiple et al., 1983; Kou et al., 2011). The influence of sulfur content on spot weld formation in stainless steel is shown in Figure 2.24. Increasing sulfur from 42 ppm (0.0042 %) to 140 ppm (0.014%) drastically increased penetration by reversing the fluid flow from outward along the surface to downward along the center of the fusion zone. The mechanism is based on the change in surface tension caused by surface-active elements, such as sulfur (Heiple & Roper, 1982; Sindo Kou, 2003) (Figure 2.25).

When vaporization is high enough to induce a depression in the weld pool, convective fluid flows are less influential in defining the resultant weld shape. Heiple et al. reported that surface-active elements showed no or negligible

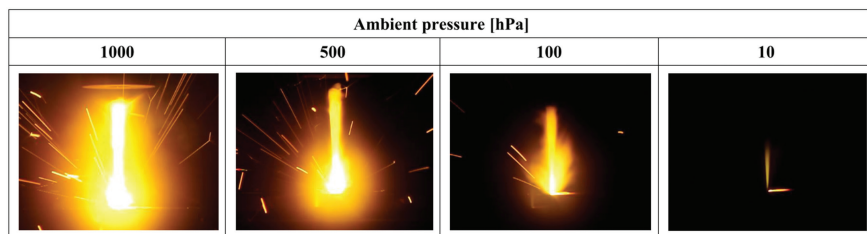


Figure 2.24 Plume formation during reduced pressure laser beam welding on S355 steel using 6000 W and traveling at 33.3 mm/s (Börner et al., 2013).

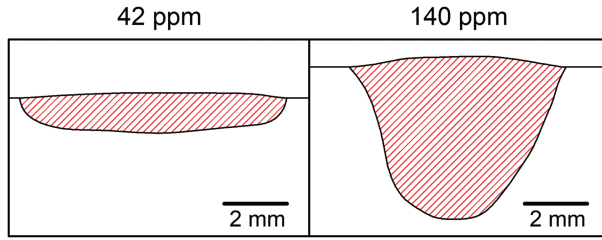


Figure 2.25 Influence of surface tension for two sulfur contents in 304 stainless steel. The schematic shows transverse fusion zone sections with 42 ppm sulfur (left) and 140 ppm sulfur (right). Results extracted from Kou et al. (2011).

influence in changing the weld profile for keyhole mode welds (Heiple et al., 1983). At the keyhole front wall, the liquid must be rapidly accelerated to divert around the vapor cavity, and these forces likely overcome surface tension and buoyancy forces. Fluid flow velocity is thought to increase with increased travel speed as a result of a change in evaporation at the keyhole front wall (Eriksson et al., 2013; Tenner et al., 2015). Keyhole welding also results in turbulent fluid flows as evident via a high propensity for porosity formation (Dowden, 2009; Fujinaga et al., 2000; Seiji Katayama et al., 1997; Matsunawa et al., 1998).

Attempts at computing respective fluid flow behavior, specifically for modeling weld behavior, are done using the Prandtl number. The Prandtl number is dimensionless and is based on the material properties for viscosity and thermal diffusivity as shown in Equation 2.23. Properly computing Prandtl numbers is based on experimentally measuring dynamic viscosity, kinematic viscosity, thermal diffusivity, and/or thermal conductivity, and/or specific heat for liquids. These values are generally unknown or poorly understood, so estimates are used, which result in substantial errors during computation. Due to the complexities in the process and the numerous interactions dictating vaporization and fluid flow, mechanisms are difficult to quantify. Models can elucidate the behavior but are often dependent on using the estimates for property values and rely on gross assumptions to generate a result.

$$P_r = \frac{\nu}{\alpha} = \frac{\mu C_p}{k} [\text{unitless}] \quad \text{Equation 2.23}$$

where P_r = Prandtl number, ν = kinematic viscosity, α = thermal diffusivity, μ = dynamic viscosity, k = thermal conductivity, and C_p = specific heat.

Summary

LBW has numerous advantages over other welding processes in applications that require reduced heat input, rapid joining times, precision, noncontact, and/or autogenous welding. After the photons are absorbed and have transferred energy into the workpiece, convective weld pool currents, heat conduction, vapor pressures, and phase transformations dictate the weld shape and penetration. The process regarding photon interactions with the solid, the liquid, the plume, and the plasma remains an active area of research due to the current, insufficient understanding.

This review provided details regarding the laser welding variables, influence of the variables on weld formation, and the physical interactions of lasers with metals and metallic alloys. Laser welding is a complex process that involves multiple states of matter and is highly influenced by material properties and process parameters. Higher power, reduced beam diameter, and slower travel speed will result in deeper weld penetration. It is well documented that the conduction-to-keyhole mode transition occurs over a parameter range and is not a finite transition. Utilizing proper shielding gas and cross-jets can also increase penetration and reduce weld discontinuities. Material compositions are reported to change the vaporization potential and fluid flow characteristics and must be considered when joining alloys with large composition ranges. For greater detail on each topic, the reader is referred to the references cited throughout this chapter.

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Chapter 3

A Review of Laser Welding Metallurgy

T. Patterson¹

¹INL – Idaho National Laboratory, Energy and Environment Science and Technology,
Idaho Falls, Idaho, United States of America

Abstract

The low heat input associated with laser beam welding (LBW) can generate relatively small fusion zones and rapid solidification rates. Because the grain structure is a result of heterogeneous nucleation and epitaxial growth along the fusion boundary, the weld pool geometry directly influences the localized solidification rate. Deep penetration welds are more susceptible to fusion zone microstructural changes as compared to elliptically shaped conduction mode welds. Higher solidification rates reduce dendrite/cell spacing and decrease elemental partitioning, which is dependent on the dendrite tip temperatures and crystallographic effects. The low heat input associated with laser welds can lead to reduced grain size and narrow heat-affected zones (HAZs), but may result in undesired phases. The high solidification rates and steep thermal gradients can also create issues such as solidification cracking. Options to alleviate cracking and alter phase transformations include modifying beam energy distributions, changing pulsing parameters, preheating, or utilizing mechanical beam oscillation across the weld pool. Therefore, understanding the solidification phenomena and weldability-related issues reported in prior literature is essential for successful LBW. This chapter reviews the fusion zone and HAZ microstructures in metallic alloys and dissimilar metal combinations and summarizes phase formation, grain size, metallurgical issues, and defect formation specific to LBW.

Keywords: Laser beam welding, Microstructure, Aluminum, Copper, Nickel, Steel, Stainless steel, Titanium, Dissimilar metals

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Introduction

A microstructure-related benefit of laser beam welding (LBW) in comparison to arc-based fusion welding processes is low overall heat input. Low heat input allows for reduced heat-affected zone (HAZ) widths for minimal influence on the area surrounding the joint and reduced residual stress formation. However, the properties proving to be beneficial may also prove detrimental. For metals, the rapid cooling rates (low heat input) can result in metallurgical issues such as high hardness phases, e.g., martensite, or insufficient phase formation, e.g., Widmanstätten austenite. The nature of LBW processes can result in variations in solidification rates, and these solidification rates influence weldability-related issues such as phase fractions, solidification cracking, and precipitation. Thus, it is necessary to understand the fundamentals of the liquid–solid transformation behavior for understanding the resultant fusion zone microstructures and laser weldability.

This review will be in relation to LBW metals and metal alloys. The goal of this review is to provide a basis for the weldability of commonly welded metallic alloy systems and how the metallic alloys are influenced by the LBW process. This includes information on microstructure discontinuities/defects and possible alleviation methods. For information regarding laser welding of polymers, the reader is referenced to Abed et al. (2018), Acherjee (2021), Ghorbel et al. (2009), and Grewell et al. (2003).

Solidification Fundamentals

During laser welding, the growth of a solid crystal or grain occurs heterogeneously, or with preferential nucleation from a solid substrate. The heterogeneous nucleation occurs because the activation barrier is smaller than that associated with homogeneous nucleation within the liquid weld pool, i.e., energetically more favorable (Porter & Easterling, 1992). The fusion boundary acts as the solid substrate where growth initiates epitaxially or with the same structure and orientation as the grain at the solid–liquid interface. The nucleated crystal continues growth in preferential growth directions, parallel to and in the opposite direction of the thermal gradient (David & Vitek, 1989; Porter & Easterling, 1992). Body-centered cubic (bcc) and face-centered cubic (fcc) metals attempt to maintain preferential growth in the $\langle 100 \rangle$ crystallographic or “easy growth” direction (Porter & Easterling, 1992; Sindo Kou, 2003). For hexagonally close-packed metals, the preferential growth direction is in the $\langle 10\bar{1}0 \rangle$ direction (Sindo Kou, 2003). Because the thermal gradient changes

depending on the location along the solid–liquid interface, grains alter their growth direction to preferentially grow perpendicular relative to the heat source isotherms. An illustration of this relative to a plan view, or top-down orientation, is shown in Figure 3.1.

The schematic shows a teardrop-shaped weld pool where the fusion boundary or isotherms maintain the same angle from the fusion boundary to the weld centerline. In this condition, there is no driving force for the grains to alter direction, but when the weld pool is rounded or takes on non-uniform shapes, the grains must alter direction to follow the maximum temperature gradient and maintain a $\langle 100 \rangle$ direction (Porter & Easterling, 1992). The grains that cannot effectively change direction become extinguished by surrounding grains that are favorably oriented to both, the thermal gradient and easy growth direction. This mechanism is referred to as “competitive” growth (Lippold, 2015; Sindo Kou, 2003).

A solid crystal forming by growth into a liquid is defined as a dendrite. The term dendrite is derived from the Greek word *dendron*, meaning tree, and in reference to the tree-shaped morphology (Kurz & Fisher, 1986). The morphology of the crystalline growth is dependent on the thermal gradient (G_L) and the solidification rate (R_L) at the solid–liquid interface (David & Vitek, 1989; Lippold, 1994). A schematic showing how the thermal gradient and solidification rate influence crystal formation is shown in Figure 3.2, with the morphologies represented in Figure 3.3. Laser welding is capable of substantially increasing the thermal gradients and solidification rate as compared to arc-based processes. The product of the thermal gradient and the solidification rate ($G_L \cdot R$) influences the scale of the solidification structure, while the ratio of the thermal gradient to the solidification rate (G_L / R) determines the solidification mode (David & Vitek, 1989; Rai et al., 2007; Sindo Kou, 2003). Generally, for metals, LB weld structures range from cellular to columnar dendritic morphologies.

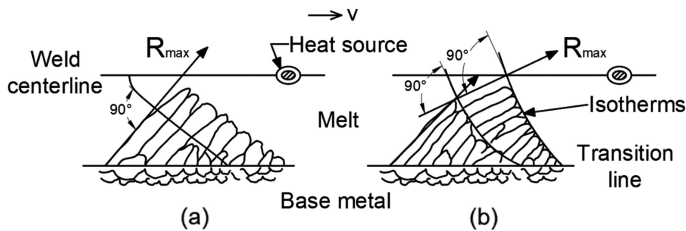


Figure 3.1 Schematic of columnar grain growth continually changing to maintain a $\langle 100 \rangle$ growth direction and perpendicular to the isotherm boundary. Recreated from Porter and Easterling (Porter & Easterling, 1992).

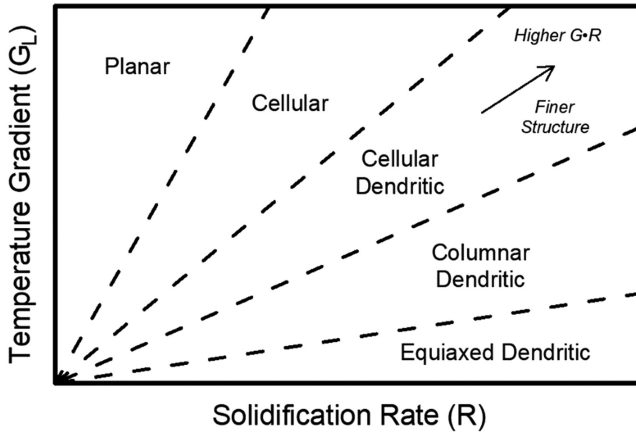


Figure 3.2 Influence of thermal gradient (G_L) versus solidification rate (R) on the crystal morphology. Recreated from Lippold (2015).

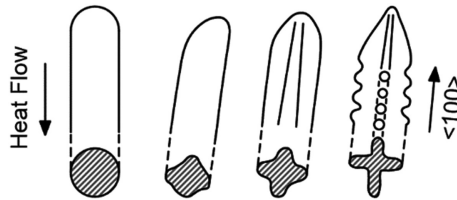


Figure 3.3 Schematic representing the transition from cellular to columnar dendritic growth (left to right). Recreated from David and Vitek (1989).

Alloyed systems melt over a range of temperatures where solid and liquid can coexist due to elemental partitioning. This temperature range defined by the solidus and liquidus in the equilibrium phase diagram is referred to as the solidification temperature range (STR). In addition to the thermal gradient and solidification rate, the STR will influence the solidification behavior by promoting constitutional supercooling (David & Vitek, 1989). These modes are shown in Figure 3.4.

Higher constitutional supercooling occurs when the actual temperature gradient is shallower than the equilibrium profile, i.e., where the remaining liquid during cooling is below the liquidus temperature. This region or “mushy zone” is composed of both solid and liquid (Sindo Kou, 2003). At high constitutional supercooling, equiaxed dendritic formation may be possible where it becomes energetically favorable to nucleate within the liquid. Suppressing constitutional supercooling to produce planar growth requires steep temperature gradients

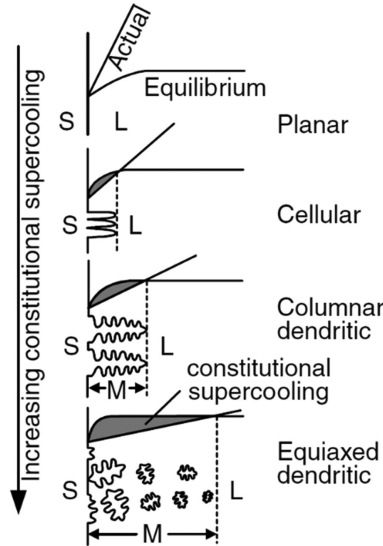


Figure 3.4 Influence of constitutional supercooling on solidification modes where S, L, and M represent solid, liquid, and mushy zones, respectively. Recreated from Kou (2003).

and low solidification rates that typically only occur at the fusion boundary. David and Vitek (1989) defined the criterion for supercooling and planar front instability as shown in Equation 3.1 (David & Vitek, 1989; Sindo Kou, 2003; Trivedi et al., 2003). This is based on the slope of the solidus and liquidus lines and partition coefficient, k , which is the composition of the solid divided by the composition of the liquid based on a nominal, binary composition at a defined temperature. The partition coefficient is shown in Equation 3.2. Therefore, to avoid the breakdown at the interface and create planar solidification, the thermal gradient must be equal to or greater than $(R_L \cdot \Delta T) / D_L$. For a more detailed explanation, the reader is referred to references (David & Vitek, 1989; Kurz & Fisher, 1986; Lippold, 2015; Sindo Kou, 2003; Trivedi et al., 2003).

$$\frac{G_L}{R} < \frac{m_L C_o (k-1)}{k D_L} \text{ or } \frac{G_L}{R} < \frac{\Delta T_o}{D_L} \left[\frac{mm^2}{s} \right] \quad \text{Equation 3.1}$$

where G_L = thermal gradient in the liquid, R = solidification rate, m_L = slope of equilibrium liquidus line, C_o = nominal alloy composition, D_L = diffusion coefficient of solute in the liquid, k = partition coefficient, and ΔT_o = temperature between the liquidus and solidus.

$$k = \frac{C_S}{C_L} [\text{unitless}] \quad \text{Equation 3.2}$$

where k = partition coefficient, C_S = composition of the solid, and C_L = composition of the liquid.

Kou presented a schematic of varying pool shapes based on heat input and welding speed as shown in Figure 3.5. At low heat inputs or travel speed, elliptical weld pools result in low constitutional supercooling and promote columnar or cellular growth. At high heat input and high welding speeds, constitutional supercooling increases with the potential for nucleation of equiaxed dendritic grains. The constitutional supercooling is based on partitioning or segregation of alloying elements into inner dendritic regions during solidification. If back diffusion of the element is not possible during cooling in the solid state, it is likely that the segregation will remain in the room temperature microstructure.

An example of epitaxial, or directional, growth from the solid–liquid interface in 304 stainless steel is shown in Figure 3.6. The overall growth direction was nearly normal to the fusion boundary at the plate surface. The slender, dark/black regions represent grain boundaries where preferential chemical segregation occurred during solidification. The dark appearance is from selective attack via electrochemical etching. The speckled appearance is

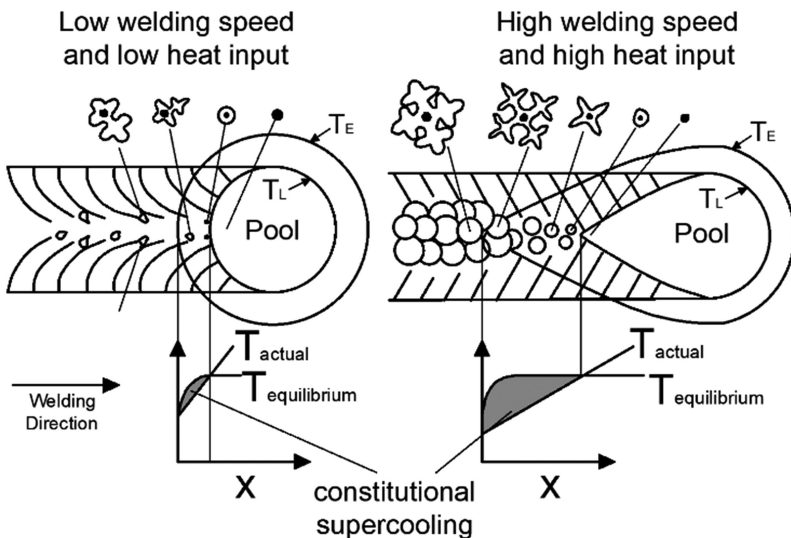


Figure 3.5 Schematic of constitutional supercooling based on heat input and weld pool geometry. Recreated from Sindou Kou (2003).

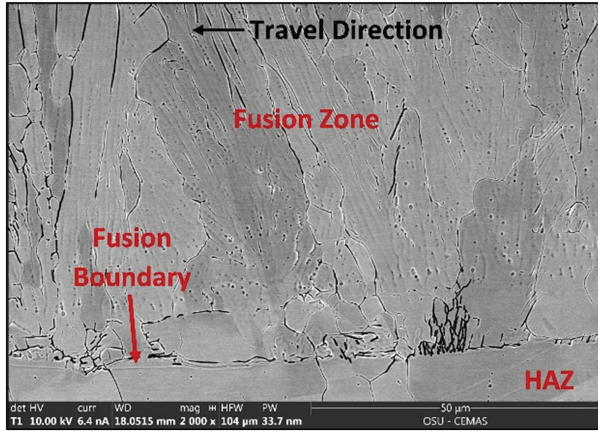


Figure 3.6 Plan view orientation (looking down on the weld) near the plate surface in a 304L stainless steel laser weld showing epitaxial grain nucleation and growth.

also a result of chemical segregation along solidification of sub-grain boundaries during the cellular dendritic solidification mode. The regions of varying shades of gray represent individual crystals (grains) in the room temperature microstructure and are elongated with respect to the growth direction.

Conduction mode welds with elliptical pools behave like arc welds at relatively low travel speeds. Figure 3.7 represents a conduction mode pool geometry and crystal growth from Rappaz et al. (1989, 1990). The welds produced in conduction mode will result in cylindrical or elliptical-shaped molten weld pools. With relatively low growth rates, grains favorably orient with the curvature of fusion boundary. However, transitioning into deep penetration LBW results in developing weld pool geometries with abrupt changes throughout the weld depths. Due to the geometric variation of the solid–liquid interface with respect to the travel direction, it is possible to develop abrupt changes in the crystal growth directions. A schematic representing a high aspect ratio weld pool and the changing growth directions is shown in Figure 3.7b. The schematics in Figure 3.7 were based on experimentally observed growth directions on an Fe-15Cr-15Ni single-crystal alloy (David et al., 1995; Rappaz et al., 1989). These single-crystal alloys may not directly represent polycrystalline materials, but nonetheless, represent the possibility of changing growth directions.

Solidification rates are determined based on the orientation of the solid–liquid interface relative to the travel velocity. For a simplistic two-dimensional approximation, the calculation is generally depicted based on the weld pool surface as in Figure 3.1. The calculation is presented in Equation 3.3 and is based on the angle of the growth direction relative to the travel direction

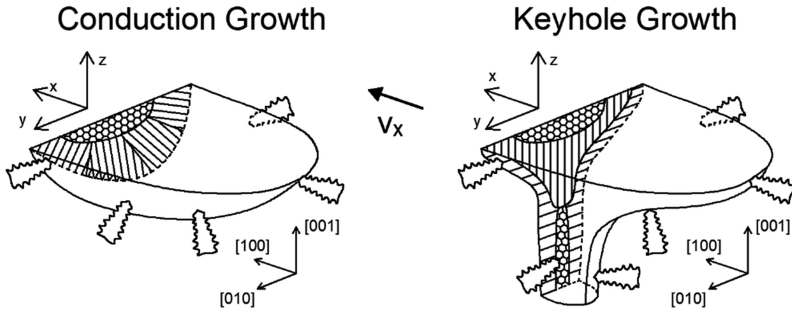


Figure 3.7 Illustration of crystal growth in an elliptical weld pool geometry as shown on the left (Rappaz et al., 1989, 1990). Illustration of keyhole weld pool showing potential preferential growth directions as shown on the right. Recreated from David et al. (1995) and Rappaz et al. (1989).

(Sindo Kou, 2003). Crystals normal to the interface ($\bar{n}$) grow at the slowest velocity and require the lowest degree of undercooling. Thus, these crystals conform to the shape of the weld pool and grow until impinged upon by more favorably oriented grains.

$$|\bar{v}_s| = |\bar{v}_t| \cdot \bar{n}_s = |\bar{v}_t| \cos(\theta) \quad \text{Equation 3.3}$$

where $\bar{v}_s$ = solidification rate, $\bar{v}_t$ = travel speed, $\bar{n}$ = unit direction normal to solid–liquid interface, and θ = angle between $\bar{n}$ and the travel direction.

For keyhole weld geometries (Figure 3.7), sharp changes to the growth directions are possible. At the transition from the keyhole to the “nail head”, grains grow toward the pool surface as in the conduction mode pool geometry. Within the keyhole finger or root region, grain growth is inward from the HAZ. Along the weld centerline and within the keyhole finger, crystal growth remains normal to the solid–liquid interface and parallel to the plate surface (Figure 3.7). However, the slender keyhole region or finger of the molten pool does not occur under all process parameters. Patterson et al. estimated the weld pool shape based on longitudinal metallographic sections along the weld centerline at weld ends or beam extinction points (Tate Patterson et al., 2022). Two of the traced solid–liquid boundaries are shown in Figure 3.8. These boundaries were revealed through a disruption of grain growth via a rapidly cooled weld pool when the beam was extinguished while the travel velocity remained constant. With a keyhole weld at 50 mm/s, the difference in solidification rate from the root to the weld surface is apparent by an inflection point near the center of depth. For the inflection point at the weld centerline, the solidification rate approached the travel speed. Doubling the travel speed

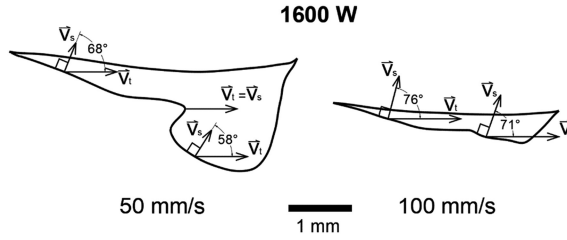


Figure 3.8 Schematic of longitudinal weld pool shapes in 304L stainless steel at two different travel speeds (Tate Patterson et al., 2022).

eliminated the inflection point and created a gradual transition from the root to the pool surface. The significance of this showed that solidification rates can be higher along the weld centerline for slower travel speeds in deep penetration LB welds.

Directional growth of dendrites into the weld pool results in parallel aligned crystals with distinct spacings relating to the thermal gradient and solidification rate. The product of these two variables results in the cooling rate as shown in Equation 3.4. The spacing between the cores or “trunks” of the dendrites is referred to as the primary dendrite arm spacing (λ_1). Similarly, a secondary dendrite arm spacing can be defined for columnar dendritic and equiaxed dendritic growth modes (λ_2). The influential factor regarding crystallographic growth into the liquid is based on the dendrite tip temperature (Trivedi et al., 2003), and this localized undercooling not only dictates the stability and growth modes but also the size of cells or dendrites. The growth of the cell or dendrite occurs due to the breakdown of planar solidification and via a perturbation advancing into the liquid. As this perturbation tip advances in alloys, the solute is rejected laterally outward and accumulates during continued growth or ripening (Kurz & Fisher, 1986; Porter & Easterling, 1992). The solute partitioning allows for distinguishing the individual cells and characterizing the dendrite arm spacing.

$$CR = G_L \cdot R \left[^\circ C / s \right] \quad \text{Equation 3.4}$$

where CR = cooling rate, G_L = thermal gradient in the liquid, and R = solidification rate.

Prior literature has reported correlations between dendrite arm spacing and cooling rates. Kou (2003) reported that the cell or dendrite arm spacing is based on a heat input to travel speed ratio and expressed a secondary dendrite arm spacing relationship as shown in Equation 3.5. This relationship is dependent on solidification time or cooling rate. Using an approximation of solidification

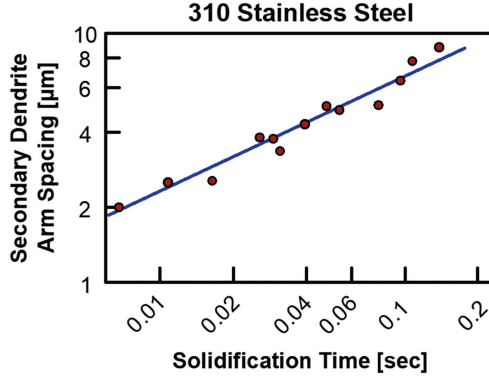


Figure 3.9 Relationship between solidification time and secondary dendrite arm spacing for 310 stainless steel. Recreated from Kou and Le (1982).

time, Kou and Le (1982) plotted a logarithmic relationship between secondary dendrite arm spacing and solidification time for 310 stainless steel as shown in Figure 3.9.

$$\lambda_2 = at_s^n = b(\varepsilon)^{-n} \quad [m] \quad \text{Equation 3.5}$$

where λ_2 = secondary dendrite arm spacing, t_s = local solidification time, a = constant, b = constant, n = constant, and ε = cooling rate.

Katayama and Matsunawa (1984) experimentally determined the constants for a Type 310 stainless steel as presented in Equation 3.6. These equations were used by Elmer et al. (n.d.) to approximate solidification rates in other stainless steel alloys. Contrary to Equation 3.6, Kurz and Fisher (1986) stated that primary dendrites do not obey these simplified relationships. Based on the results of Trivedi (1970, 1980), Kurz and Fisher (1986) reported that the primary arm spacing is proportional to the difference in temperatures from the dendrite tip and root divided by the mean thermal gradient. The calculation is shown in Equation 3.7 (Kurz & Fisher, 1986).

$$\lambda_1 = 80(\varepsilon)^{-0.33}, \lambda_2 = 25(\varepsilon)^{-0.28}, \quad [m] \quad \text{Equation 3.6}$$

where λ_1 = primary dendrite arm spacing, λ_2 = secondary dendrite arm spacing, and ε = cooling rate.

$$\lambda_1 = \sqrt{\frac{3(T^* - T'_s)R_L}{G_L}} \quad [m] \quad \text{Equation 3.7}$$

where λ_1 = primary dendrite arm spacing, T^* = dendrite tip temperature, T_s' = eutectic or solidus temperature, R_L = solidification rate, and G_L = thermal gradient.

After solidification, the cooling rates will determine further grain growth, precipitation formation, and phase transformations. The influence of the cooling rate is dependent on the alloy properties and process parameters, but the crystallographic structure will maintain remnants of solidification structures via segregation, elongated grains, crystallographic texture, or prior phase boundaries, e.g., prior austenite or prior beta grain boundaries. The following sections review LBW microstructures for the fusion and HAZs of metallic alloys and dissimilar metal welds reported in the literature.

Materials

Weld quality and microstructure changes are directly related to the metal alloys, alloying elements, and impurity contents. The following sections will address laser weld microstructures and microstructure-related discontinuities associated with aluminum alloys, nickel alloys, steel alloys, stainless steel alloys, titanium alloys, and dissimilar metal laser welds.

Aluminum Alloys

Common applications for welding aluminum alloys include automotive, marine, aerospace, and defense industries. The first time the aircraft industry adopted laser welding was for the aircraft fuselage at the turn of the twentieth century (Cao et al., 2003; Ion, 2000). Similarly, laser welding of tailor-welded blanks for the automotive industry has been underway for over two decades (Ion, 2000). Aluminum alloys are approximately three times lighter than steel alloys and about 1.5 times lighter than titanium alloys. Although the yield strengths are lower than steel or titanium alloys, aluminum alloys are advantageous where high strengths are not needed and costs are a major design factor. Aluminum alloys are categorized as non-heat treatable and heat treatable. The non-heat-treatable solid-solution alloys are pure aluminum, aluminum–manganese, aluminum–silicon, and aluminum–magnesium designated as 1000, 3000, 4000, and 5000 series, respectively. The precipitation-strengthened alloys include aluminum–copper, aluminum–magnesium–silicon, and aluminum–zinc–magnesium and are designated as 2000, 6000, and 7000, respectively (Ion, 2000). The 8000 series classifies other alloys, but a common

addition is lithium. Alloying with lithium can reduce the density by 3% with up to a 6% increase in the elastic modulus (Ion, 2000).

Challenges associated with LBW aluminum include high reflectivity, higher thermal conductivity, low boiling point elements, high coefficient of thermal expansion, large STR, large solidification shrinkage, low melting temperature constituents, solubility of hydrogen in liquid, and tenacious oxide layers. These properties allow for liquation, solidification cracking, and porosity, which increase the difficulty in creating defect-free fusion welds. If solidification-related issues can be overcome during fusion welding, the region surrounding the weld will be influenced by phase dissolution, precipitate coarsening, recrystallization, and/or grain growth. The heating influence in a conduction-like mode weld is shown via the binary phase diagram schematic in Figure 3.10.

An example of solidification cracks formed from autogenous, i.e., no filler metal, pulsed laser welds is shown in Figure 3.11. These solidification cracks were generated in an aluminum–copper alloy, 2024. Filler metals are likely needed to eliminate solidification cracking when laser welding heat-treatable alloys such as 6061. The necessity in utilizing a filler material is to increase the amount of low melting temperature eutectics. Without enough eutectic liquid, cracks formed from solidification shrinkage stress cannot be effectively backfilled or healed. Filler metals commonly used to promote eutectic healing include 4043, 4047, and 4145, which are high silicon-containing alloys that promote the low melting temperature aluminum–silicon eutectic (Liu &

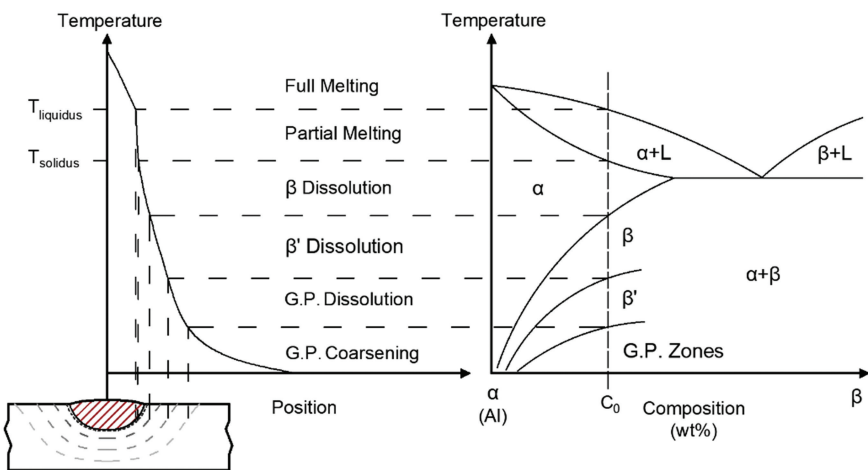


Figure 3.10 Schematic diagram showing the regions of the heat-affected zone in a binary aluminum alloy.

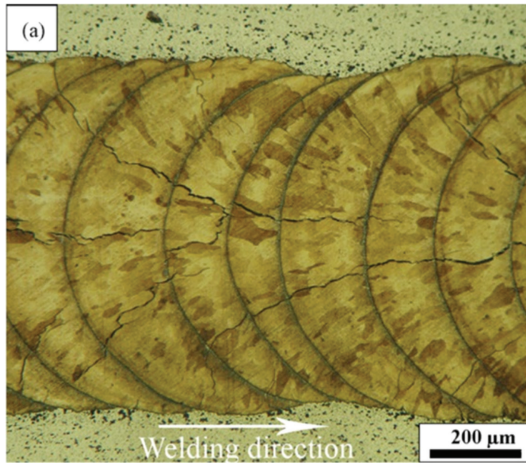


Figure 3.11 Solidification cracks in pulsed laser welds on aluminum alloy 2024 with an O temper (Ghaini et al., 2009).

Kou, 2017). Junai et al. showed that silicon contents greater than 4 wt% will produce crack-free fusion zones during pulsed LBW (Junai et al., 1990). Other filler metal alloys reported to be beneficial include 5000 series alloys, such as 5183. A 5183 as a filler metal was shown to promote the formation of equiaxed grains. The zirconium particles acted as nucleation sites, which stifled the directional growth of long, parallel columnar grains that would contain liquid films and higher stresses across their boundaries. By decreasing the grain size, the stresses and liquid films are reduced through distribution over a greater grain boundary area (Binroth et al., 1990).

In addition to solidification cracking, aluminum alloys may also be susceptible to liquation cracking, either in the HAZ or in reheat weld metal. An example of HAZ liquation is shown in Figure 3.12. The liquation is evident via the dark lines penetrating from the fusion zone into the partially melted region of the HAZ. It is the segregation of alloying elements and low melting temperature phases that allow preferential melting of the grain boundaries and flow of liquid further from the fusion zone. The right photomicrograph in Figure 3.12 shows a higher magnification of the partially melted zone, which is revealed by the presence of liquated particles within the grains and continuous films along the grain boundaries. However, Pakdil et al. (2011) reported that cracks were not detected despite the grain boundary liquation. An example of where the grain boundary liquation and intrinsic stresses resulted in HAZ liquation cracking is shown in Figure 3.13. These welds were deposited in a 7000 series alloy. Various mechanisms are theorized to cause HAZ liquation,

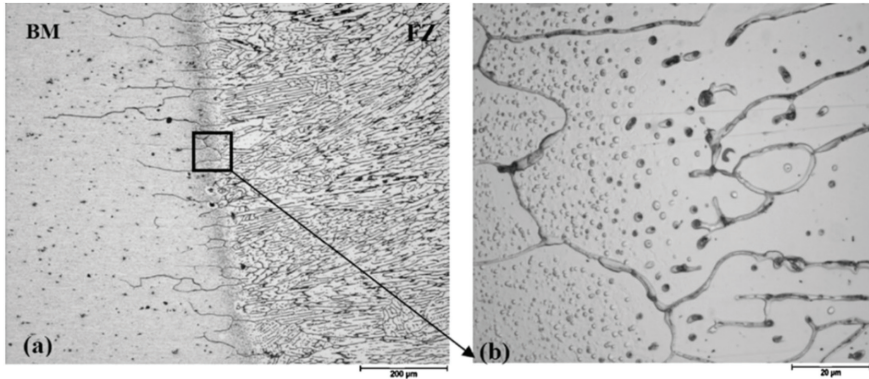


Figure 3.12 Laser weld on aluminum alloy 6056 showing the partially melted zone and liquation along grain boundaries (Pakdil et al., 2011).

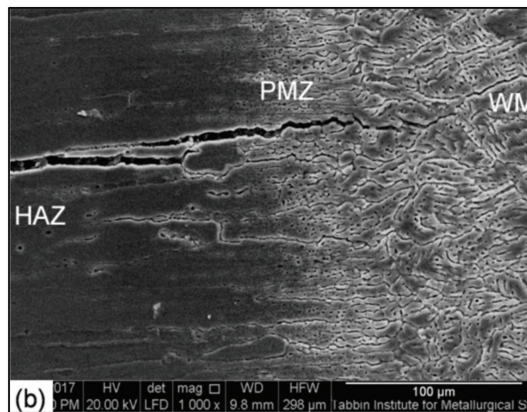


Figure 3.13 Microstructure of laser weld in aluminum alloy 7075-T6 showing cracking from the fusion zone (right) into the heat-affected zone (El-Batahgy et al., 2021).

such as constitutional liquation or pipeline diffusion. These mechanisms are outside the scope of this chapter, but further information can be found in the following reference (Lippold, 2015).

Some aluminum alloys are nearly un-weldable using fusion welding processes. This includes the 7000 series alloys, such as 7075. An example of issues occurring during laser welding of 7075 in the T6 heat-treated condition (solutionized and artificially aged) is shown in Figure 3.13 (El-Batahgy et al., 2021). These alloys are selected for applications where light-weighting and high base material strength are a necessity. Therefore, filler metals that are generally used for welding other alloy systems to avoid solidification cracking,

such as 4043, result in reduced strength properties. Also, LBW is an inherently low heat input process with respect to fusion welding. Thus, it is difficult to further reduce heat input to avoid HAZ liquation. As with hot cracking, methods employed to avoid liquation include reduced heat input, welding in the solutionized condition, and reducing extrinsic stresses.

Nickel Alloys

Nickel alloys are largely used in applications involving corrosive fluids, cryogenic temperatures, or elevated temperatures. Some of these industries include aerospace, power generation, pulp and paper, chemical processing, and petrochemicals. The two main categories of nickel-base alloys used in industrial applications are categorized as solid-solution strengthened or precipitation strengthened. Other classifications include pure nickel or oxide dispersion strengthened alloys.

Solid-solution strengthening elements include Co, Cr, Cu, Fe, Mo, Ta, and W. Additions of Nb, Ti, W, Mo, and Ta will form MC-type carbides, which can promote grain boundary liquation in the partially melted zone (David et al., 2015). Precipitation-strengthened alloys utilize the formation of gamma prime (γ') and gamma double-prime precipitates (γ''). Gamma prime is a precipitate formed from aluminum/titanium additions ($\text{Ni}_3\text{Al/Ti}$), and gamma double prime is a nickel niobium precipitate (Ni_3Nb). Typically, material strength reduces with increased temperature, but the elevated temperature precipitation results in the unique nature of high temperature strength – often referred to as superalloys.

Laser weldability of nickel-based alloys can be inferred from arc welding analysis. As with the austenitic stainless steels, nickel-based alloys have higher susceptibility to solidification cracking as a result of austenite (fcc) solidification, low thermal conductivity, large STR, and high viscosity in the liquid phase. Alloys highly susceptible to solidification cracking via arc welding, are likely more susceptible during LBW. Pakniat et al. showed that Hastelloy® X could be welded with a continuous wave laser, but solidification cracks were unavoidable with pulsed LBW. However, increasing pulsed frequency showed a reduction in the total crack length (Pakniat et al., 2016). Laser welding investigations on a Ni-Cr-Fe alloy, Inconel® Alloy 690, showed that multiple solidification structures formed within a single keyhole laser weld, which is indicative of the varying thermal gradient and solidification rate. The change in growth modes was related to the solidification growth rate based on weld pool shape computed through a mathematical model that was used to generate a solidification map (Blecher et al., 2014).

HAZ microstructures in nickel-based alloys can undergo particle dissolution, grain growth, grain boundary segregation, and precipitate formation. It is likely for grain growth to only be observed with high heat inputs in solid-solution strengthened alloys (Allen et al., 2015). Multi-component systems using alloying elements that readily form MC-type carbides and Laves phases may be susceptible to HAZ liquation cracking. The segregation or diffusion of these elements along the grain boundaries result in large STRs due to low melting temperature eutectics. Temperatures above the solidus line allow for particles to liquate and propagate liquid along the grain boundaries. Upon cooling, the thermal strains exhaust the strength of the liquid, forming intergranular cracking. Liquation cracking has been reported in both solid solution and precipitation-strengthened alloys. Often this form of cracking is reported as microcracking or microfissuring due to the small cracks that typically span the length of one to a few grain boundaries (Allen et al., 2015). However, liquation cracks should not be confused with ductility dip cracking, which have also been referred to as microfissuring, and is an entirely different cracking mechanism.

Ren et al. showed liquation cracks in laser welds in the solid-solution strengthened Ni-Cr-Co-Mo alloy, Inconel® 617 (Ren et al., 2015). Precipitation-strengthened alloys may be highly susceptible to liquation cracking because of the addition elements that readily form MC-type carbides and the Laves phase. Figure 3.14 shows a HAZ liquation crack near the keyhole finger of an autogenous laser weld in Inconel® 740H®. Similarly, HAZ liquation cracking in Haynes® 282, a competitive alloy to 740H® laser welds, was shown by Osoba and Ojo (2012). Although it is generally regarded that low heat input, such as in LBW, reduces HAZ liquation cracking susceptibility (DuPont et al.,

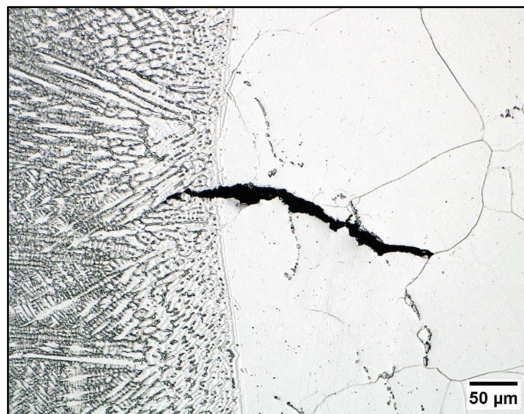


Figure 3.14 HAZ liquation crack in laser-welded Inconel® 740H®.

2009), the open literature has reported HAZ liquation cracking in a variety of laser-welded nickel alloys. For laser welds, reports have indicated that higher heat input can reduce or eliminate HAZ liquation cracking (Idowu et al., 2007; Osoba & Ojo, 2012). Idowu et al. (2007) argued that the higher heat input welds increased the liquid film for greater crack healing, and the shallower thermal gradients reduced stresses across the boundaries.

Steel Alloys

Steel alloys are used in a wide range of applications including automotive, agriculture, and marine industries. Automotive applications use advanced high-strength steels, such as the dual-phase or high-strength low alloy steels, for maintaining strength while reducing the overall vehicle weight (Saha et al., 2014). The high-strength potentials for steel alloys allow for reducing material thickness to decrease product costs and reduce material usage, and the manufacturing process generates fewer carbon dioxide emissions than aluminum alloys. Because lasers can be easily automated and can rapidly bond components, they are advantageous for rapid manufacturing of steel components. Heat inputs are also advantageous for joining thin materials for vehicle body panels and agricultural harvesters.

Some challenges with LBW steel alloys include high hardness, embrittling microstructures such as martensite or lath bainite. Steel alloys with carbon concentrations below approximately 0.17 wt% are likely to solidify as bcc or ferrite, which is less susceptible to solidification cracking (Lippold, 2015). Within the HAZ, transformation will depend on the alloy content with an example of the iron-carbon (iron-carbide) diagram shown in Figure 3.15.

Assuming a carbon greater than the peritectic point (δ -ferrite+liquid $\leftrightarrow\gamma$ -austenite), the HAZ will transform into austenite, or fcc, as the temperature rises above 721°C (A_{c1}). The coarse grain heat-affected zone (CGHAZ) will undergo full transformation into austenite, with sufficient time for large grain growth, hence, coarse grain. In theory, the region adjacent to the CHGAZ will transform into austenite, but with reduced time and temperature, grain growth is reduced, which results in the fine grain heat-affected zone (FGHAZ). Because most laser beam welds are made with low input, it is likely that it will not be possible to discern the difference between the CGHAZ and FGHAZ. Next, the full transformation into austenite will be a two-phase region consisting of partial transformation to austenite on heating, known as the partially transformed zone. Because the stable room temperature phase is ferrite, the austenite will transform to a more energetically stable state. The rapid cooling rates do not allow for diffusion-based transformations to occur, resulting in

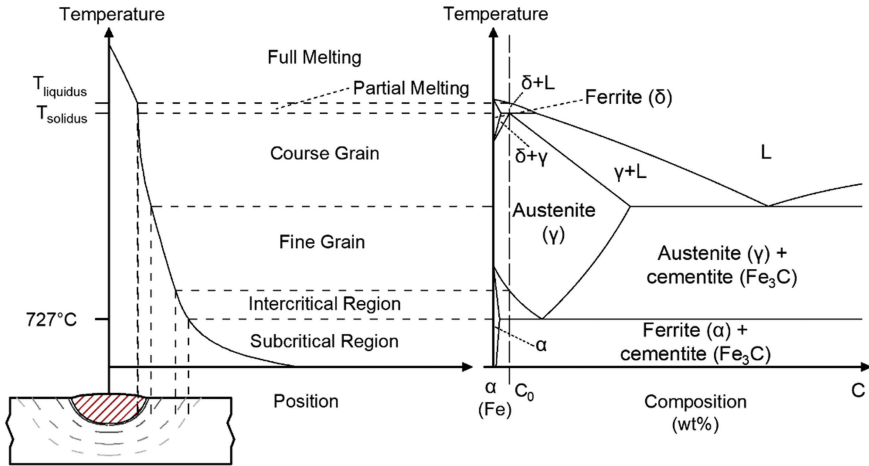


Figure 3.15 Schematic of heat-affected zone microstructure formation in steel.

a metastable martensite, or the body-centered tetragonal (bct) phase. Strictly considering iron and carbon, the as-welded room temperature structure will be martensite, ferrite, and cementite carbides. For the region heated below the A_{c1} temperature, phase transformation to austenite is not possible, but grain growth, precipitation, and carbide coarsening will result.

Low carbon steels that solidify as delta-ferrite prior to solid-state transformation to austenite are generally less susceptible to solidification cracking. However, solidification cracking is still possible for low carbon steels where deep penetration welds induce high amounts of thermally induced intrinsic stress. Higher carbon steels solidifying as austenite are susceptible to solidification cracking. Extrinsic stresses and solidification rates should be minimized to reduce the chances of forming solidification cracks.

The biggest issue regarding laser welding of low alloy steels is embrittlement with high hardness microstructures within the fusion zone and HAZ. Fresh untampered martensite will form and reduce the ductility of the welded joint. Although there are no known reports relating hydrogen embrittlement to laser welds, without proper hydrogen control it is possible to introduce hydrogen to cause hydrogen-induced cracking. To mitigate high hardness martensite or bainitic structures, preheating or post-weld heat treating is necessary to either slow the cooling rate or temper the martensitic structure.

Stainless Steel Alloys

Highly alloyed steels with chromium contents above approximately 10.5–11.5 wt% are considered stainless steel and used in applications requiring corrosion

resistance (Lippold & Kotecki, n.d.). These applications include paper mills, food processing, nuclear energy, and defense systems. Stainless steels are further classified into austenitic, ferritic, duplex, precipitation hardened, and martensitic subclasses, each with varying proportions of chromium, nickel, and additional alloying elements.

The most common stainless steel alloy is Type 304, which is an austenitic grade consisting of 18 wt% chromium and 8 wt% nickel. Further classifications are 304L and 304H, where L and H stand for low (≤ 0.03 wt%) and high (0.04–0.1 wt%) carbon, respectively. Forming ferrite on solidification is advantageous when welding austenitic stainless steels by reducing the susceptibility to solidification cracking (Lippold, 1985; Lippold & Kotecki, 2005; Lippold & Savage, 1979; Vitek et al., 1983). The transformation into ferrite or austenite for a given set of process conditions is based on the proportions of elements added to the alloy. Applicable elements that thermodynamically stabilize ferrite include Cr, Mo, Nb, Ti, and Si, and applicable elements that stabilize austenite include Ni, C, N, Mn, and Cu. Assuming a ternary alloy that only contains Fe, Cr, and Ni, pseudo-binary phase diagrams were developed by Lippold and Savage (1979). These represent the phase formations at varying SS compositions under equilibrium cooling conditions. Figure 3.16 shows similar diagrams from Kou and Le (1982).

The pseudo-binary diagrams show the differences in phase formations by altering the chromium/nickel contents while iron remains constant. A triangular region exists showing the coexistence of ferrite, austenite, and liquid. Compositions left of this triangle stabilize primary austenite (Type A) solidification, while compositions right of this triangle stabilize primary ferrite (Type F) solidification. Within this three-phase region, solidification can occur as primary austenite with secondary ferrite (Type AF) or primary ferrite with secondary austenite (Type FA).

The complexity associated with alloys containing a variety of elemental compositions skews the distinct boundaries depicted through pseudo-binary phase diagrams, and software programs are required for computing multi-element systems. In an effort to predict ferrite content in stainless steel weld metal, numerous diagrams were developed from experimental arc welding data (Kotecki & Siewert, 1992; Long & Delong, n.d.; Schaeffler, 1949). These diagrams weight the percentages of alloying elements in the analysis.

Using this data, “maps” of solidification modes were produced according to relationships based on chromium-to-nickel equivalencies ($\text{Cr}_{\text{eq}}/\text{Ni}_{\text{eq}}$) relating to ferrite and austenite promoting elements, respectively. A common diagram is the WRC-1992 diagram from Kotecki and Siewert (1992) and is reported to be the most accurate and reliable predictive diagram for austenitic and duplex stainless steel weld metal (Lippold & Kotecki, 2005). The formulas for three equivalency relationships are shown in Equation 3.8–Equation 3.10

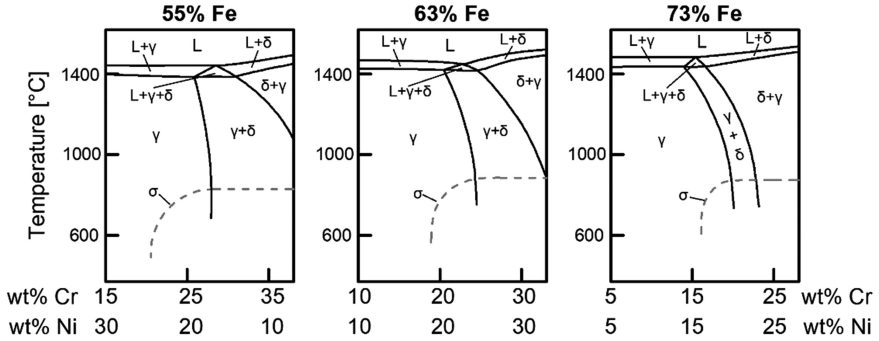


Figure 3.16 Pseudo-binary Fe–Cr–Ni phase diagrams from Kou and Le at 55% Fe (left), 63% Fe (center), and 73% Fe (right). Recreated from S. Kou and Le (1982) and Sindo Kou (2003).

from Kotecki and Siewert (WRC-1992), Suutala, and Hammar and Svennson, respectively (Hammar & Svennson, 1979; Kotecki & Siewert, n.d.; Kujanpää et al., 1979). However, the diagrams are based on arc welding conditions and do not predict LBW microstructure formation. Therefore, additional diagrams or microstructure “maps” were updated for LBW solidification and cooling rates using the previously developed equations.

Equation 3.8: WRC-1992 Cr_{eq} and Ni_{eq} formulas (Kotecki & Siewert, 1992)

$$Cr_{eq} = Cr + Mo + 0.7Nb$$

$$Ni_{eq} = Ni + 35C + 20N + 0.25Cu$$

Equation 3.9: Suutala Cr_{eq} and Ni_{eq} formulas (Kujanpää et al., 1979)

$$Cr_{eq} = Cr + 1.37Mo + 1.5Si + 2Nb + 3Ti$$

$$Ni_{eq} = Ni + 0.31Mn + 22C + 14.2N + Cu$$

Equation 3.10: Hammar and Svennson Cr_{eq} and Ni_{eq} formulas (Hammar & Svennson, 1979)

$$Cr_{eq} = Cr + 1.37Mo$$

$$Ni_{eq} = Ni + 0.31Mn + 22C + 14.2N$$

The change in solidification mode along with restraint, grain orientation, and impurity levels can lead to increased cracking susceptibility during LBW.

Because transitioning from primary ferrite to primary austenite solidification through increased solidification rates along the weld pool may result in solidification cracking, empirically developed diagrams are useful for selecting alloys for avoiding solidification cracking (Lippold, 1994). Pulsed LBW is known to exacerbate solidification cracking based on the rapid solidification. Chelladurai et al. (2015) reported that cracks in Type 316 were avoided by welding in keyhole mode, with low heat input conduction and transition mode welds resulting in solidification cracking. To extend the stainless steel laser welding microstructure map developed by Lippold (1994) and to perform the analysis for continuous wave LBW, Patterson et al. produced laser beam welds at a range of heat inputs with alloy chromium–nickel equivalencies ranging from Type 304L to 2205 duplex stainless steel (Tate Patterson et al., 2022). The updated microstructure map is shown in Figure 3.17, and examples of optical micrographs and electron backscattered diffraction data used to generate the map are shown in Figure 3.18.

The 304L solidification structure shows a vermicular or skeletal ferrite because of elemental partitioning, indicative of primary ferrite and secondary austenite solidification. However, the EBSD map shows mostly austenite with small regions of ferrite that do not pertain to the elemental partitioning on solidification because most of the nucleated ferrite was transformed into austenite during cooling in the solid state. On the contrary, the 2205 duplex stainless steel solidified as ferrite and cooling rates were fast enough that Widmanstätten austenite was not able to transform along the ferrite grain boundaries. The ability to maintain ideal ferrite/austenite ratios is an issue when LBW duplex stainless steels because the austenite levels below 50%

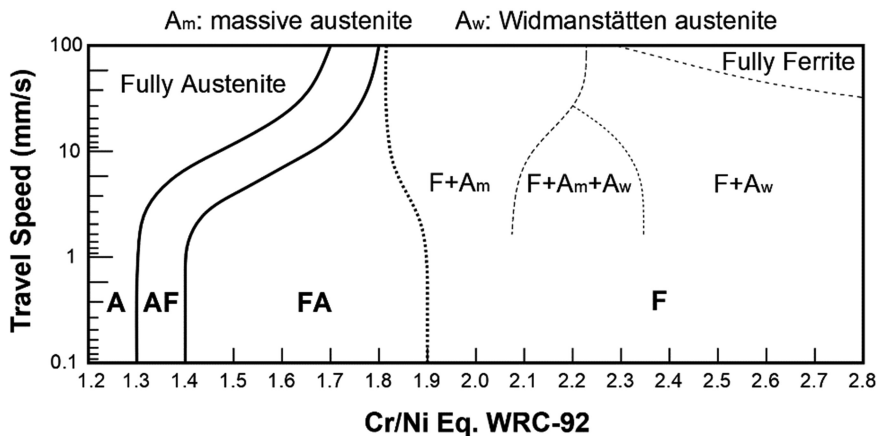


Figure 3.17 Updated LBW stainless steel microstructure map of travel speed versus WRC-1992 Cr_{eq}/Ni_{eq} (Tate Patterson et al., 2022).

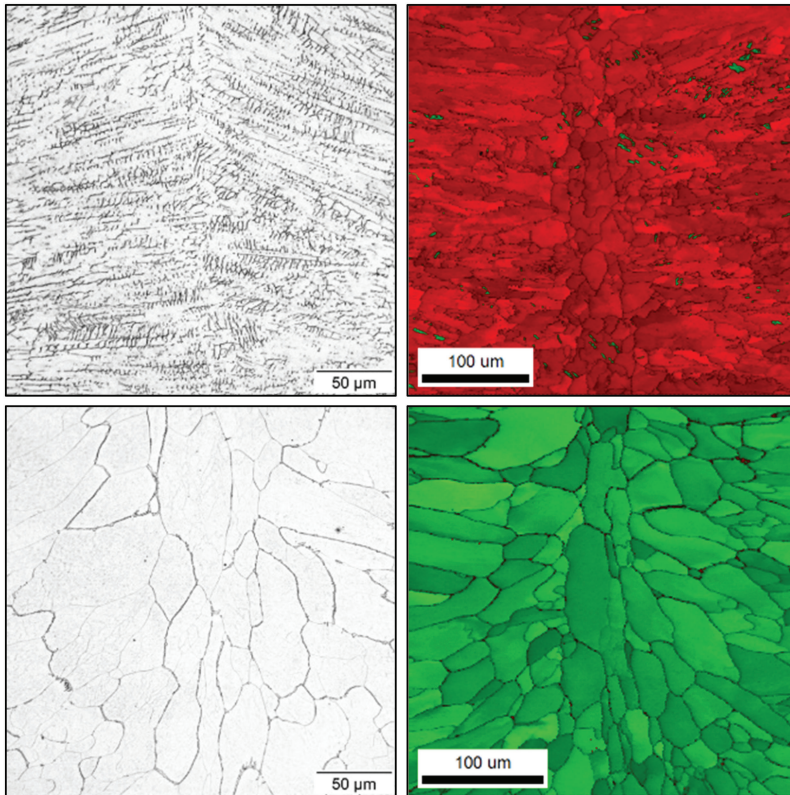


Figure 3.18 Comparison of optical micrographs (left) and electron backscattered diffraction scans (right) showing fractions of austenite (red/dark) and ferrite (green/light). Weld metal structures in the as-welded condition are shown for 304L stainless steel (top) and 2205 duplex stainless steel (bottom) (Tate Patterson et al., 2022).

degrade corrosion performance and reduce mechanical properties (Karlsson, 2012; Muthupandi et al., 2003; Sieurin & Sandström, 2006; Verma & Taiwade, 2017). Nitrogen is a potent austenite stabilizing element and only second to carbon as shown in Equations 3.9 and 3.10. LBW will vaporize the nitrogen from the weld pool resulting in reduced nitrogen in the fusion zone and along the fusion boundary in the HAZ.

Titanium Alloys

Lasing welding of titanium alloys is common in aerospace applications, and the most used alloy is Ti-6Al-4V. Other applications include medical devices

such as orthopedic and prosthetic implants, sports equipment such as golf clubs, and petrochemical components. Alloys are designated as alpha, near-alpha, alpha-beta, and beta alloys. Pure titanium is allotropic with the high-temperature solid phase as body-centered cubic (bcc) and the low-temperature stable phase as hexagonal close packed (hcp) (Gammon et al., 2018). An example of fusion zone microstructures for pure titanium produced via continuous wave LBW is shown in Figure 3.19. The structures consist entirely of alpha, and the prior beta grains that formed on solidification are nearly indiscernible.

Titanium alloy additions stabilize either alpha or beta phases at room temperature. Alpha stabilizers include elements such as Al, C, N, O, and Sn. Beta-stabilizing elements fall into one of two categories, which include isomorphous additions and eutectoid additions. The isomorphous beta stabilizers include Mo, Nb, Re, Ta, W, and V. Eutectoid beta stabilizers include Cr, Co, Cu, H, Fe, Mn, and Ni. An example of the transformations that occur in the HAZ for increasing the content of an isomorphous beta-stabilizing element is shown in Figure 3.20, and an example of a eutectoid forming element is shown in Figure 3.21.

The challenges associated with LBW of titanium alloys are largely associated with oxidation, porosity, martensite formation, and beta grain growth. The use of proper shielding gas setups or welding in a vacuum environment will remove oxidation (Mazumder & Steen, 1982). However, porosity is difficult to eliminate and is exacerbated by deep penetration welds. Small pores or “micropores” less than 0.25 mm in diameter have been reported to result from hydrogen gas absorbed into the weld pool (Auwal et al., 2018) or atmospheric

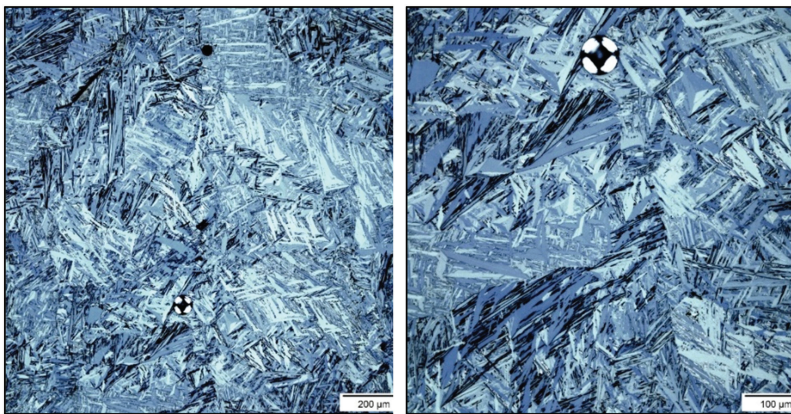


Figure 3.19 Laser weld fusion zone microstructure in pure titanium showing alpha grains. Kroll’s etch. Immersed for ≈ 10 s. Captured using differential image contrast (DIC) (Tate Patterson, 2021).

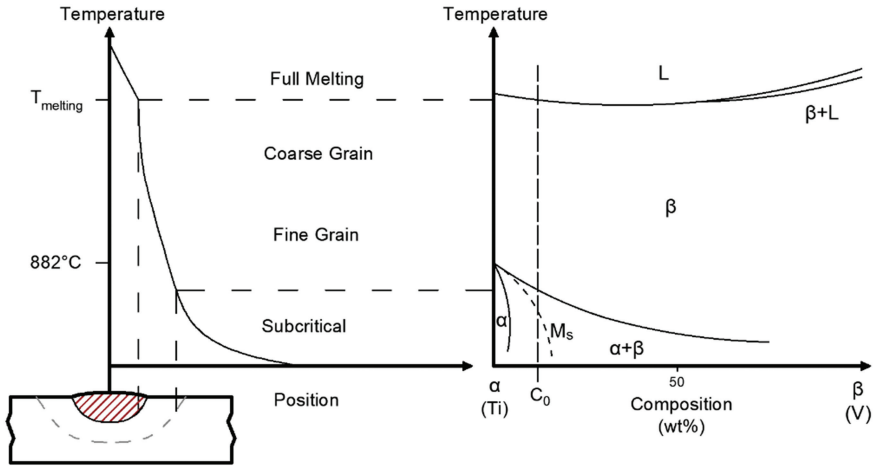


Figure 3.20 Schematic of titanium alloy heat-affected zone regions with an isomorphous beta-stabilizing element in titanium.

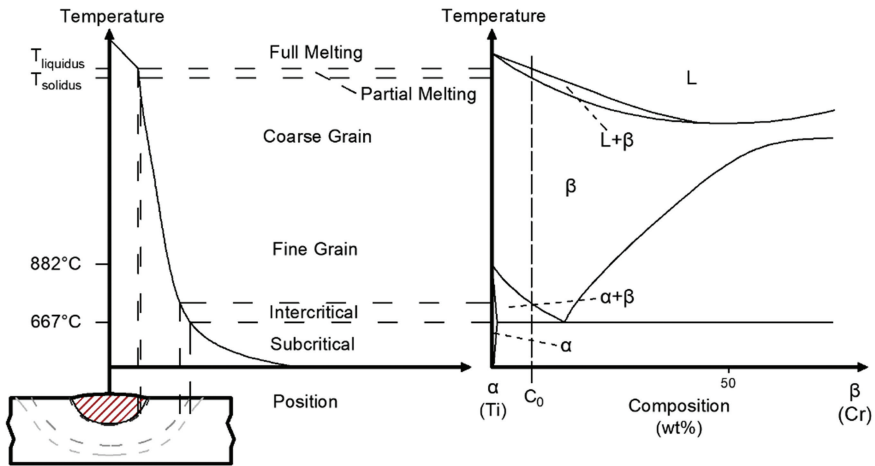


Figure 3.21 Schematic of heat-affected zone microstructure with a eutectoid beta-stabilizing element in titanium.

gas entrapment (Kumar & Sinha, 2019), and Khaled (1994) reported porosity as potential crack nucleation locations. The rapid cooling rates during LBW can result in the distorted hexagonal close-packed structure or α' -martensite formation in near-alpha and alpha-beta alloys. The martensitic structure can result in reduced weld metal and HAZ ductility.

Grain growth in the high-temperature beta phase is difficult to control in titanium alloys. The beta grains epitaxially nucleate from the fusion boundary and rapidly grow inward toward the weld centerline until impinged upon by growth from the opposite fusion boundary. The rapid diffusion rates in the bcc structure reduce elemental segregation during solidification, which nearly eliminates solidification cracking and remnants of solidification structures (Baeslack et al., 1993). The main concern with large grain sizes is reduced ductility in the welded joint (Baeslack & Banas, n.d.; Richter et al., 2007). Low LBW heat input, compared to arc welding, has the greatest potential to reduce grain size and minimize lengths of the α' martensite. Utilizing parent grain reconstruction algorithms from electron backscattered diffraction data, Patterson et al. reconstructed prior beta grains in Ti-6Al-4V laser welds as shown in Figure 3.22. By controlling the weld deposit size, grain sizes were reduced as depicted in Figure 3.22 (Patterson et al., 2022).

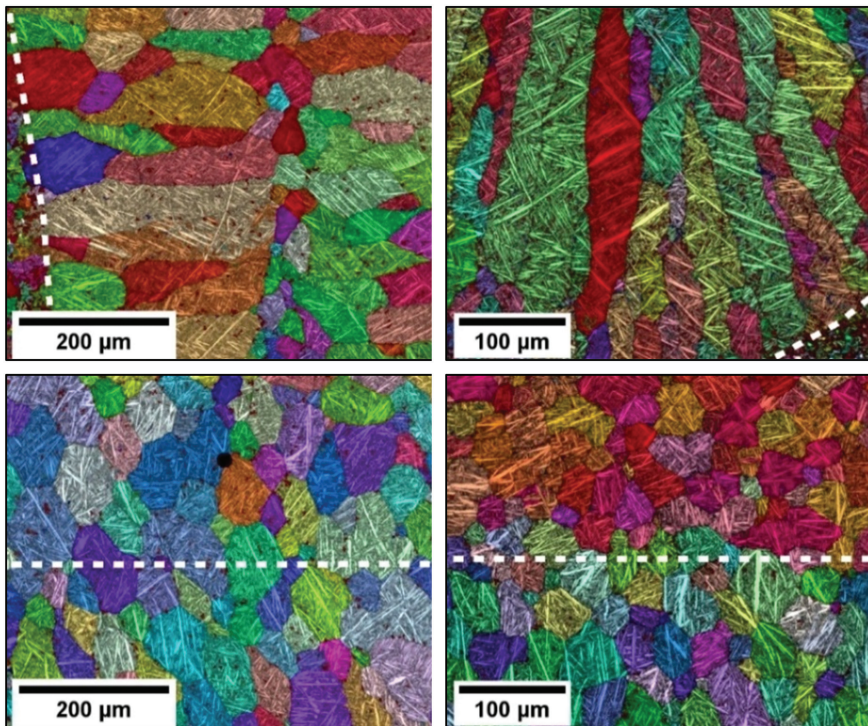


Figure 3.22 Beta grains in Ti-6Al-4V laser beam welds at 100 mm/s (left) and 50 mm/s (right) (T. Patterson et al., 2022).

Dissimilar Metal Welds

Joining of dissimilar alloys can be particularly advantageous for fabricating components with specific properties in explicit regions. The properties are generally for improved strength, decreased weight, corrosion resistance, or electrical/thermal conductivity. This section describes some of the common dissimilar metal welds reported in the open literature. Significant challenges exist when bonding dissimilar metal combinations, and the intended application must be critically analyzed to avoid joint failure from weld-related cracks, brittle intermetallic compounds, galvanic corrosion, and so on. Metallurgical compatibility between dissimilar metals can be based on the Hume–Rothery rules relative to substitutional or interstitial alloying elements. The elemental miscibility is based on atomic size, crystal structure, chemical affinity, and valencies (Sun & Ion, 1995). The larger the difference in chemical and physical properties of the two metals to be joined, the greater the joining difficulty. In addition to the metals, non-metallic alloying additions must also be considered to account for detrimental precipitate formation. A table showing the approximate laser weldability of dissimilar combinations for pure metals is shown in Table 3.1 (Sun & Ion, 1995). The ability to generate suitable welds is defined as a letter scale, with fair and poor scales representing difficult combinations or combinations that should be avoided. Although depicted as compatible, it should not be assumed that successful welds can be produced over large parameter ranges or without preheating, post-weld heat treating, oxide removal, and so on.

At the time of this writing, the automotive industry remains invested in dissimilar metal joining to reduce overall vehicle weight and improve efficiency. With dissimilar metal combinations, vehicles can be designed with regions of materials having high strength and toughness held together using lightweight alloys at less critical interconnections. An example of this is joining high-strength, dual-phase steels, such as Usibor® or Ductibor® alloys, to aluminum alloys. However, iron and aluminum will form brittle intermetallic compounds, as shown in Figure 3.23, which severely degrade the structural integrity of the joint. Additionally, the differences in thermal properties result in drastically different melting temperatures. In addition, the coefficient of thermal expansion mismatches the result in high residual stresses, which exacerbate solidification cracking (Chen et al., 2011; Ma et al., 2014). Cracks are a result of brittle, intermetallic phases and large material property mismatches.

Other joints attractive to the automotive industry include dissimilar metal welds between copper and aluminum, typically in a lap joint configuration. These joints are commonly used in battery terminal connections for electric vehicles. Jarwitz et al. evaluated aluminum to copper laser welds with

Table 3.1 Dissimilar metal laser weldability

	W	Ta	Mo	Cr	Co	Ti	Be	Fe	Pt	Ni	Pd	Cu	Au	Ag	Mg	Al	Zn	Cd	Pb
Ta	E																		
Mo	E	E																	
Cr	E	P	E																
Co	F	P	F	G															
Ti	F	E	E	G	F														
Be	P	P	P	P	F	P													
Fe	F	F	G	E	E	F	F												
Pt	G	F	G	G	E	F	P	G											
Ni	F	G	F	G	E	F	F	G	E										
Pd	F	G	G	G	E	F	F	G	E	E									
Cu	P	P	P	P	F	F	F	F	E	E	E								
Au	*	*	P	F	P	F	F	F	E	E	E	E							
Ag	P	P	P	P	P	F	P	P	F	P	E	F	E						
Mg	P	*	P	P	P	P	P	P	P	P	P	F	F	F					
Al	P	P	P	P	F	F	P	P	P	F	P	F	F	F	F				
Zn	P	*	P	P	F	P	P	F	P	F	F	G	F	G	P	F			
Cd	*	*	*	P	P	P	*	P	F	F	F	P	F	G	E	P	P		
Pb	P	*	P	P	P	P	*	P	P	P	P	P	P	P	P	P	P	P	
Sn	P	P	P	P	P	P	P	P	F	P	F	P	F	F	P	P	P	P	F

E = excellent, G = good, F = fair, P = poor, * = no data.

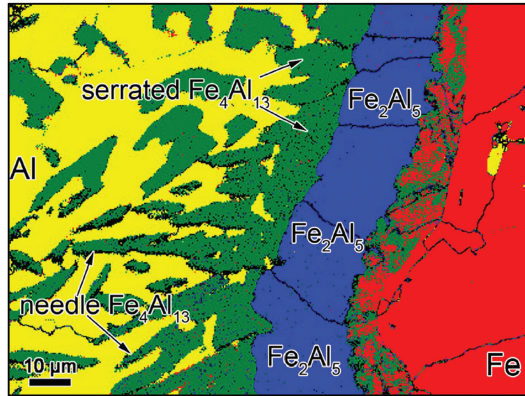


Figure 3.23 Steel to aluminum laser weld in a square groove, butt joint configuration showing different phase formation and intermetallic compounds between aluminum and steel (Cui et al., 2017).

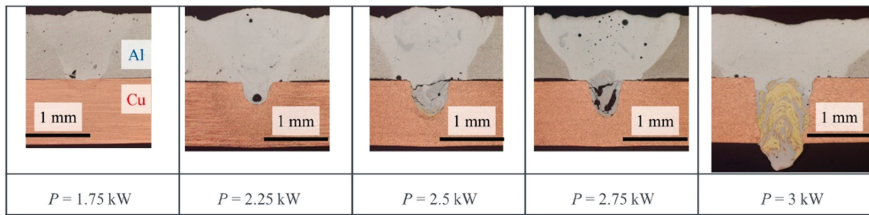


Figure 3.24 Aluminum to copper laser welds (Jarwitz et al., 2018).

increasing power, as shown in Figure 3.24 (Jarwitz et al., 2018). Low weld penetration into the copper resulted in cracking through the fusion zone near the Al–Cu interface. Mathivanan and Plapper attempted to use beam oscillation and pulsing to minimize porosity and disperse the intermetallic Al–Cu compounds. Although many of the reported joints show porosity and a large area fraction of intermetallic formation due to high intermixing between the copper and aluminum (Mathivanan & Plapper, 2019). Huang et al. (2020) showed the distribution and differences in mixing relative to laser power and travel speed. Welds where overall copper concentrations exceeded nearly 60% in the aluminum plate region resulted in fusion zone cracking.

As with steel and aluminum, steel and titanium or stainless steel and titanium joints are largely incompatible, which is caused by limited solid solubility of iron and titanium. Intermetallic compounds include bcc structures of FeTi and Fe₂Ti, and if it is a stainless steel, TiCr₂, NiTi₂, and Ti₅Fe₁₇Cr₅ (Quazi et al., 2020; Shanmugarajan & Padmanabham, 2012). These compounds will greatly

reduce the joint strength, and the thermal expansion mismatches between both alloys will cause to high internal stress. Titanium to aluminum is also difficult to fusion weld based on the metallurgical incompatibility of aluminum and titanium. Some of the easily formed intermetallic compounds include TiAl , Ti_3Al , and TiAl_2 (S. Chen et al., 2016).

Dissimilar metal combinations do not only include joining of two different metals and include joining of similar alloys. Some common examples of these dissimilar combinations include welding different steels, stainless steels, aluminum alloys, and titanium alloys to an alloy within their respective alloy class. The aircraft industry can benefit from dissimilar metal welded connections such as commercially pure titanium aircraft skin to Ti-6Al-4V webbing (Froend et al., 2017). Similarly, automotive applications can benefit from aluminum alloy combinations, such as welds between alloy 5052 and 6061, which Ramiarison et al. (2022) showed successful joints when using beam oscillation. Other aluminum alloy combinations reported in literature include 5000 series to 7000 series and 2000 series to 6000 series alloys (Ion, 2000).

Dissimilar welds on stainless steels are commonly performed and can be aided with the use of the Schaeffler diagram modified for laser welding processes (Schaeffler, 1949). This modified Schaeffler diagram is shown in Figure 3.25 (Sun & Ion, 1995). By computing the chromium and nickel equivalencies of each stainless steel alloy to be joined, the fraction of austenite, ferrite, and/or martensite can be approximated. The importance of this is to avoid solidification cracking by maintaining nucleation of primary ferrite.

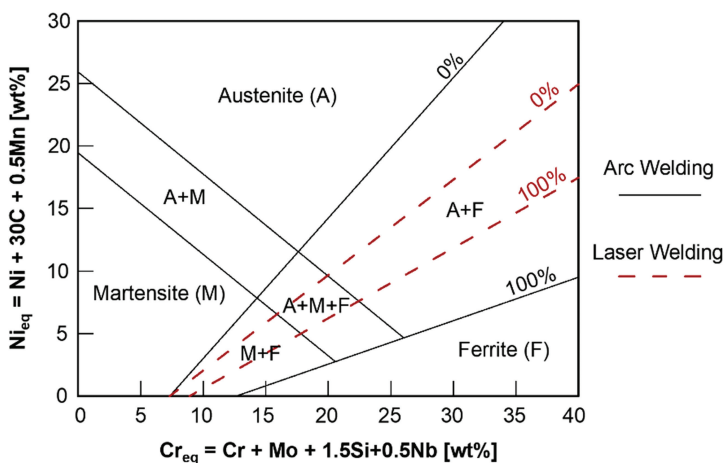


Figure 3.25 Modified Schaeffler diagram to account for rapid solidification rates in laser welding. Recreated from Sun and Ion (1995).

Generally, welding similar alloy groups together result in fewer weldability issues because of greater metallurgical compatibility. Issues can include higher levels of alloying elements in the fusion zone or diffusion of elements from the weld metal into the HAZ resulting in embrittlement. If drastically different alloys are to be welded that do not form solid solutions or do not show excellent weldability, the best approach for generating structurally sound bonds is to utilize metallic interlayers between the incompatible materials. These interlayers must be chosen such that the interlayer is compatible with both alloys, and welds must be deposited such that the interlayer maintains a barrier between the incompatible materials to be joined. The use of interlayers such as Mg, Cu, Nb, Ta, and V has been reported to improve weldment strength in certain alloy combinations (Gao et al., 2012; Quazi et al., 2020). This review only covers a few of the laser-welded dissimilar metal combinations reported in literature. Many combinations, interlayers, and parameter studies have been published regarding dissimilar metal joining, and the references cited in this review can provide some additional information on this topic.

Summary

LBW can be an advantageous fusion welding process that provides reduced heat input with a wide range of possible penetration depths. Metallurgically, LBW is capable of minimizing grain growth, reducing time for precipitation of undesirable phases, or minimizing the over-aging of precipitates. Table 3.2 shows a synopsis of different alloys and microstructure-related discontinuities. The relative relations provided in Table 3.2 are not an exact representation of the issues associated with each alloy group but they are a broad overview and generally hold true. For alleviating some of these issues during welding, Table 3.3 summarizes the general rules regarding laser and parameter selection. These are general strategies and will not apply in all laser welding cases or alloys, and every weld must be analyzed on a case-by-case basis.

As laser beam processes continue to evolve, parameters allow for greater flexibility in manipulating microstructure in alloys traditionally difficult to fusion weld. Although this review only covered fusion welding microstructures, lasers are used in other welding applications such as laser hybrid welding, laser impact welding (Wang et al., 2020), or laser brazing/laser braze welding (Ogura et al., 2016; Saida et al., 2006). These processes extend laser welding capabilities and can help to further mitigate the weldability issues addressed in this chapter. As this review is not an exhaustive analysis of laser welding microstructures, the readers are encouraged to review the provided references for more detail.

Table 3.2 Relative laser weldability issues/susceptibilities for metal alloys rated as low, moderate, and high

Alloys	Reflectivity	Oxidation	Solidification cracking	Liquation cracking	Porosity	Grain growth
Aluminum	High	High	High	High	High	Low
Copper	High	Low	High	Moderate	Moderate	Moderate
Nickel	Moderate	Moderate	High	High	Moderate	Low
Steel	Moderate	Low	Low	Low	Low	Moderate
Stainless steel	Low	Moderate	Moderate	Moderate	Moderate	Low
Titanium	Moderate	High	Low	Low	Moderate	High

These are approximations and broad classifications.

Table 3.3 Laser weldability issues and corresponding methods for reducing the influence/effect

Weldability issues	Continuous wave or pulsed wave	Power density	Travel speed	Environment
Reflectivity	Pulsed	High	Low	NA
Oxidation	NA	NA	NA	Nonreactive/inert shielding gas or vacuum
Solidification cracking	Continuous	Low	Low	Oxidation free
Liquation cracking	Pulsed	NA	High	NA
Porosity	NA	Low	Low	Nonreactive/ inert shielding gas or vacuum
Excessive grain growth	Pulsed	High	High	NA

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Chapter 4

A Review on Laser Welding of Ni-Ti Based Shape Memory Alloy

S. Datta*, M. S. Raza, P. Saha, and D. K. Pratihari

Department of Mechanical Engineering, Indian Institute of Technology Kharagpur,
Kharagpur, West Bengal India-721302

Abstract

NiTinol, the functional advanced material with shape memory effect (SME), is largely used for different industrial applications, including biomedical, automotive, aerospace, and power plants. The superior functional aspects of the material, namely superelasticity (SE), SME, and biocompatibility, offer numerous advantages to design engineers. However, some complexity arises in the method of fabrication of the material and sets restrictions on the extensive utilization of the material for the manufacturing of products with complex contours. The major issues related to fabrication of the material are limited machinability and weldability of the material. The problem of weldability is ended with the formation of brittle intermetallic compounds, modification in shape memory behavior, alteration in phase transformation temperature, reduction in strength, and changes in superelastic behavior. Even, the joining of NiTinol to other materials creates a lot more problems owing to the variation in thermophysical and chemical properties of different base materials. Hence, an in-depth research in the field of welding of NiTinol is very much needed. So far many researchers studied different aspects of joining NiTinol with similar and dissimilar material combinations. This chapter presents a thorough evaluation of laser welding of NiTinol as laser welding is a cost-effective and dependable method for joining of NiTinol alloy.

* susmita.prod@gmail.com, susmitaforall@gmail.com

Keywords: Laser welding of NiTiInol; Microstructure; Thermomechanical properties; Corrosion resistance performance; Phase transformation temperature; Post-weld heat treatment

Introduction

NiTiInol and Its Importance

In order to meet the different industrial requirements, newer tailor-made materials are developed for particular applications. The invention of NiTiInol and the identification of its superior properties were one of the most significant developments of metallic alloys in the twentieth century. In the year 1959, the invention of equiatomic Ni-Ti alloy was carried out by William J. Buehler in Naval Ordinance Laboratory and named the discovery as NiTiInol, which represents both its elemental composition and the place of origin. During the subsequent development phase of the alloy, the importance of the alloy was identified by W. Buehler and Fredrick E. Wang in 1962 [Buehler et al., 1963; Kaufmann & Mayo, 1997]. Yet the full potential of NiTiInol to be utilized in different industries cannot be exploited by researchers due to the different difficulties associated with the material itself [Mehrpooya et al., 2017]. Fundamentally, NiTiInol fits into a particular class of materials, known as shape memory alloy (SMA). Recently, the widespread use of SMAs in diverse fields could be attributed to the exclusive properties, namely superelasticity (SE) and shape memory effect (SME), a state-of-the-art solution to many engineering problems. These properties are basically the consequence of solid-state phase transformation phenomena between martensite and austenite phases triggered by either stress or temperature [Elahinia et al., 2012; Hartl & Lagoudas, 2008].

NiTiInol alloys are the mostly used commercial SMAs in comparison with other types of SMAs. More than 90% market of the SMA applications is captured by NiTi-based SMAs like NiTiInol, NiTi-Cu, and NiTi-Nb [Elahinia, 2016; LExcellent, 2013]. NiTiInol can be used for multiscale applications, starting from large-scale civil structural applications to microelectromechanical (MEMS) devices. The capability to withstand large deformation, excellent corrosion and fatigue resistance, and good biocompatibility are the reasons behind its widespread use in different fields of life. Different surgical devices to be used in human physiology are designed to take the advantage of fully recoverable deformation mechanism of this kind of alloy [Yamauchi et al., 2011; Petrini et al., 2011]. With the achievements in understanding the smart and superior functionalities of the material, this alloy is widely used in cardiovascular and orthopedic surgery devices, solid-state heat engines, orthodontics,

safety products, and shrink-to-fit aircraft pipe couplers and has revolutionized the field of medical, engineering, and manufacturing industry. Along with other superior properties, nonmagnetic properties and unique properties, which match with human bones and tissues, make NiTiNol a preferred choice in biomedical industries. Because of the low stiffness value, NiTiNol is successfully used as an alternative material for bone implants [Andani et al., 2014; Amerinatanzi et al., 2016; Andani et al., 2016]. Nowadays, NiTiNol is extensively used for stents, eyeglass frames, and anchors used to attach tendons with bone and orthodontic archwires. In today's world, the alloy has become a good choice among design engineers of different fields including biomedical applications, aviation and aerospace applications, MEMS, hydrospace equipment, civil and structural applications, and others [Morgan, 2004; Song et al., 2006; Hartl & Lagoudas, 2007; Jiang & Spearing, 2012; Patil & Song, 2017; Wen et al., 2018]. Table 4.1 shows different thermophysical properties of the alloy.

Successful commercialization of any product depends on appropriate fabrication methods. Many unique products with low cost can be produced with SMAs by the development of appropriate joining methods. The fabrication of NiTiNol imposes challenges due to the poor machinability and formability of the alloy, and the lack of available joining techniques [Mehrpouya et al., 2017]. Successful implementation of any joining technique is not only defined by

Table 4.1 Different thermophysical properties of NiTiNol [Jani et al., 2014]

Property	Symbol	Units	Martensite	Austenite
Corrosion resistance	—	—	Similar to 300 series SS or Ti alloy	Similar to 300 series SS or Ti alloy
Density	ρD	kg/m ³	6450–6500	6450–6500
Electrical resistivity	ρR	$\mu\Omega\text{cm}$	76–80	82–100
Specific heat capacity	C	J/kg K	836.8	836.8
Thermal conductivity	k	W/m K	8.6–10	18
Thermal expansion coefficient	α	m/m K ⁻¹	6.69×10^{-6}	11.09×10^{-6}
Ultimate tensile strength	σ_{UTS}	MPa	895 (fully annealed)/1900 (hardened)	
Young's modulus	E	GPa	28–41	75–83
Yield strength	σ_y	MPa	70–140	195–690
Poisson's ratio	K	—	0.33	—
Magnetic susceptibility	χ	μ_{emug}	2.5	3.8

the strength of the joint but also by the changes associated with the chemical composition and microstructure of the material in the joint. Superelasticity and shape memory behavior must have considerable effects on the joining methods used [Chan et al., 2015]. Hence, the joining of NiTiNol is completely different from the joining of commonly used alloys of titanium and nickel. As a result of this, special attention in the selection of both the appropriate joining method for NiTiNol and ranges of process parameters is essential [van der Eijk et al., 2003].

Use of Laser as a Tool for Manufacturing and Laser Welding in Specific

Laser (light amplification by stimulated emission of radiation) is basically a monochromatic and coherent source of electromagnetic waves, which can be transmitted in a straight line path with insignificant divergence. Laser has widespread applications ranging from ordinary everyday use to sophisticated commercial and scientific use. It can also be used for saving the life as well as threatening the life. Laser has the capability of generating concentrated heat on solid surfaces. This heat generated by lasers can be successfully used for different types of cost-effective processing of materials such as forming, drilling, cutting, surface treatment, and welding. These methods are advantageous in comparison with their traditional counterparts because of the superiority and the novelty of the processes themselves.

Laser beam welding (LBW) is a process of joining, where the joint is obtained by melting the surfaces to be joined by the use of heat generated by the impingement of the concentrated coherent laser beam on the surfaces to be joined. The LBW process has many advantages compared to its conventional counterpart. Those include (i) high melting efficiency, (ii) high depth of penetration, (iii) low heat-affected zone (HAZ), (iv) high depth-to-width ratio, (v) high welding speed, (vi) possibility of joining thin to thick materials, (vii) capability of welding dissimilar materials, (viii) used for unreachable areas, (ix) clean welding method because of non-contact type of weld, (x) operated in the open air, (xi) transmitted over long distance with minimum loss of power, (xii) low total thermal input, and (xiii) high accuracy.

Depending upon the modes of heating, laser welding is classified into two groups. Those are conduction mode and keyhole mode of welding. Actually, the variation in the value of power density is the cause behind the variation in the mode of laser welding. Power density is described as the laser power focused on the unit area of the laser spot footprint.

The power density in the conduction mode of welding is sufficient to cause the melting of metal. The energy is transferred by the method of conduction. Basically low power density ($5 \times 10^5 \text{ W/cm}^2$) for conduction mode creates a

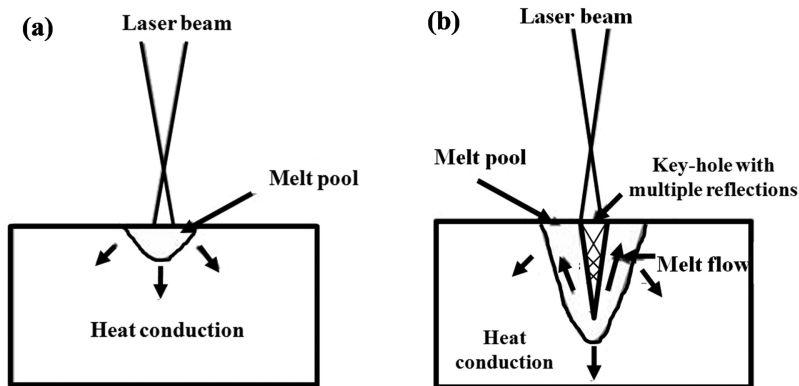


Figure 4.1 Schematic illustration of the mode of welding: (a) conduction mode and (b) keyhole mode (Steen and Mazumder, 2010).

shallow and wide weld nugget (Steen & Mazumder, 2010). The two surfaces, to be joined, melt to full thickness by conduction of heat from the top hot surface. Figure. 4.1(a) schematically represents the conduction mode of welding.

Keyhole mode of welding attains the weld depth in a special way. In the keyhole mode, the laser power density is adequately high to cause the vaporization of metal along with melting. The metal vapor generates the gas that moves outward. This basically creates the keyhole or the passageway from the top to deep inside the material. The amount of laser entered inside the keyhole is determined by the attenuation originated by the absorption of the laser in the plasma plume. The energy is captivated within the keyhole wall by the phenomena of Fresnel absorption as the laser beam is reflected several times by the keyhole wall. In this way, the keyhole plays a significant role to transfer and distribute the laser energy into the depth of the material. The surface vaporization from the keyhole wall helps to maintain the tunnel during laser welding. The vaporized material helped to keep the keyhole open by acting against the surface tension of the molten pool. A deep narrow weld will be created by the movement of the laser beam across the surface till the power density is high enough and the travel speed is not very high. Figure. 4.1(b) represents the schematic representation of the keyhole mode of welding.

Reasons behind Laser Welding of NiTiInol

Since the discovery of NiTiInol in the year 1960, several research works have been done for the upgradation of the manufacturing process without much affecting its superior properties. Conversely, there are many influences that restrict the extensive use of this material in diverse fields. Meager

machinability and deficiency of available joining methods of NiTiInol are the source of all challenges for manufacturing any element with this material. The functional behavior and mechanical responses of the alloy are significantly affected by thermal influence and chemical impurity during joining. Attaining noble quality joints in making devices and apparatuses with intricate geometry in NiTiInol by possessing its superior qualities intact is very challenging.

Different welding techniques, such as brazing with silver paste [Quintino et al., 2013], resistance spot welding [Tam et al., 2012], friction welding [Fukumoto et al., 2010], and tungsten inert gas welding [Lue et al., 2013] were attempted. Brazed-joined NiTiInol components can be utilized for micro-sized devices, where the strength requirement of the joint is not the major criterion. In friction stir welding and resistance butt-welding, trimming action is essential to make the job finish. Gas tungsten arc welding speedily diminishes the mechanical properties of the material. Laser welding is typically favored for joining of NiTiInol [Mehrpouya et al., 2018] among other fusion welding methods because of the superior features of the process, such as the capability to attain good control of heat input, extraordinary processing speed, decent weld-bead profile, thin width of HAZ, minimum residual stress, and distortion. CO₂ gas laser [Tuissi et al., 2004], Nd:YAG solid-state laser [Maletta et al., 2009], and fiber laser [Miranda et al., 2015] were used by researchers for the welding of NiTiInol SMA.

Review on Laser Welding of NiTiInol

Laser welding is a major joining technique for NiTiInol alloys among all other available joining methods because of the associated inherent advantages of the process. The superior characteristics of laser welding include narrow weld-bead and HAZ due to higher density of energy and small heat input. The reproducibility of the process and the lower thermal stress give additional advantages to the process. The use of different laser sources in the joining of NiTiInol helps in understanding the effect of welding on different mechanical and microstructural aspects. An in-depth comparison of welding of NiTiInol associated with different laser sources is not a very easy task, as different laser sources introduce different effects in the process. Even a little compositional variation in the base material, having different thermomechanical properties, furnishes different results. In this part of the study, the main focus will be the review of literature associated with laser welding of NiTiInol with various sources of laser and the consequence of different process parameters on microstructural variations, mechanical properties, phase transformation temperature, and SME.

Nd:YAG laser [Haas and Schuessler, 1999; Sevilla et al., 2008], CO₂ laser [Hirose et al., 1990], fiber laser [Chan et al., 2011; Chan & Man, 2011], and diode laser with continuous as well as pulsed mode of operation can be used to generate welded joints with different properties. Among all, fiber laser and Nd:YAG laser are the most used source of laser for welding sleek SMAs as they produce narrow HAZs with higher accuracy [Schuessler, 2011; Chan et al., 2011; Iijima et al., 2008; Pouquet et al., 2012]. As the lower value of the wavelength of these lasers helps in better absorption of the same, the overall energy used for the process is less. The pulsed mode of operation is favored for the joining of thin sheets and wires, whereas the continuous mode of operation is extensively used for jobs having more thickness with better mechanical properties [Oliveira et al., 2017; Vieira et al., 2011].

In most cases, operating parameters mostly affect the effectiveness of the manufacturing process. Laser power, scan speed, laser wavelength, the position of the focal spot, spot diameter, type of protecting gas supply, and the flow rate of the same are the major process parameters in laser welding [Gisario et al., 2017a; Mehrpouya & Emamian, 2017]. The reproducibility of the output data can be controlled by regulating laser power and scan speed. Otherwise, heat input, the intensity of power, and thermal characteristics of the material such as thermal conductivity, thermal capacity, and heat loss [Chan et al., 2012a; Gisario et al., 2017b; Gisario et al., 2016] stalwartly controls the weld-bead dimension. The microstructure of the weld-bead can be smoother due to lower heat input and rapid cooling in comparison with other welding processes. The gas shielding system basically takes care of the reaction of the molten pool with oxygen, nitrogen, and hydrogen and prevents the formation of brittle oxides and nitrides [van der Eijk et al., 2003; Ion, 2005].

So far, some researchers considered laser welding of NiTi₉₀ to NiTi₉₀. Some impressive work in that field can be described as follows. It was reported in a study (Gugel et al., 2008) that by suitable selection of process parameters, the maximum tensile strength of the YAG laser-welded 100 μ m NiTi₉₀ wire could reach up to 75% of the parent NiTi₉₀ wire, and the dimension of fusion zone (FZ) and HAZ could also be reduced. They used an average power of 100 W with Ar gas shielding system. In another study by Schlossmacher et al. (1997), the microstructural variations and mechanical properties of Nd:YAG laser-welded Ni-48.9Ti sheet were reported. They observed that the joint strength was 80% of the base material while the elongation to failure ratio was 8.4% at a welding speed of 180 mm/min with a gas shielding system, and the pulsed mode of the laser was used by them. Tuissi et al. (1999) employed Nd:YAG laser for joining of 0.54 mm NiTi₉₀ sheet (Ni-49.6 atomic%). They examined the SME and SE of different microstructural conditions. Khan and Zhou (2008) showed enhanced mechanical performance of the joint with a

high value of peak power and a lesser value of pulse frequency. In their study, Vieira et al. (2011) showed the SE behavior of continuous mode Nd:YAG laser-welded NiTiInol sheets in a cyclic loading test and reported the superior functional properties of the joint with a recoverable strain value of more than 6%. Oliveira et al. (2017) used different laser power levels varying from 790 W to 1485 W to study the functional characteristics of the laser-welded NiTiInol samples. The shape memory property was measured by a bending test. For this, the NiTiInol samples were deformed and dropped off and then they were heated over austenite finish temperature. It was observed that the weld-zone fully recovered to its initial structure. They too found that the base material was in the austenite phase at room temperature, whereas both the austenite and martensite phases coexisted in the FZ and HAZ. However, this variation did not influence the shape memory recovery property of this alloy. Gong et al. (2011) got a good quality weld, having 97% tensile strength of that of the base material and ductile mode of failure, for laser welding NiTiInol sheet of 0.2 mm thickness by the use of impulse laser technology.

Although Nd:YAG laser was vastly employed for joining of NiTiInol alloys, the use of other laser sources was also reported by many researchers. Hsu et al. (2001) used CO₂ laser source for joining of NiTiInol alloys. Hsu et al. (2001) reported augmentation in the strength of the welded samples with a lower value of elongation than the base material and the increment of B2 phase occurred in the welded samples. Chan et al. (2011) compared fiber laser welding having a small beam diameter and high intensity with Nd:YAG laser welding of 0.25-mm-thick NiTiInol foil. Full penetration weld with significant mechanical property was achieved. Even they performed post-weld heat treatment (PWHT) to pick up the corrosion resistance behavior of the welded joint. The post-process annealing stalwartly enhanced the recovery mechanism of the weld-zone. It was found that fiber laser welding was more capable in the micro-fabrication of NiTiInol alloys because of small spot diameter, higher power intensity, and increased rate of energy delivery with shorter interaction time and reduction in HAZ area [Chan and Man 2011]. Mehrpouya et al. (2018), in their recent work, examined the effect of laser welding on NiTiInol thin sheet using a high-power diode laser through microstructural property, functional behavior, and mechanical property study.

Mechanical Properties

Tensile Property

Laser welding considerably reduces the tensile property of both the melt pool and the HAZ. In the case of Ti-rich alloy, having a martensite phase at room temperature, a considerable decline in UTS (almost 30% in association with

the base material) of the joint was observed due to the occurrence of micro-cracks in the weld-bead and occurrence of brittle intermetallic compounds in the fusion region and also in heat-affected region [Tuissi et al., 2004]. Laser-welded samples had a strength near to half the value of the parent material [Maletta et al., 2009; Chan et al., 2012a]. Input parameters strongly affected the mechanical performance of the welded samples.

Khan and Zhou (2008) reported that both the highest strength and ductility of the welded samples were reduced from that of the parent material. It was also reported that a high value of energy and a low value of pulse frequency helped to get better mechanical characteristics of the joint [Song et al., 2008]. Zeng et al. (2016) examined the consequence of laser parameters, such as laser current and time on joint breaking force for SME and PE samples. The welded joints failed at higher power due to the formation of softer HAZ, and the inadequate low power caused interfacial failure. Joint breaking force (JBF) + reduces for both the cases of similar and dissimilar joining with the increase of laser duration. It happened due to rapid melting and solidification phenomena of the material of the welded region, a good location for stress concentration. If laser parameters were not chosen cautiously, it would definitely deteriorate the overall tensile property of the joint [Lavvafi, 2013].

Cyclic Deformation Behavior

Hsu et al. (2001) reported cyclic deformation characteristics of laser-welded NiTiNol plates for 50 cycles and the strain value was varied up to 3%. A high value of residual strain and the level of stress required for the formation of stress-induced martensite were reported by them. Gugel et al. (2008) investigated the fatigue performance of laser-welded NiTiNol wire for a number of cycles. A reduction in plateau stress was seen with a rise in the number of cycles and a substantial difference also occurred in the transformation of B2→B19' phase. Even a permanent deformation was also observed following ten cycles.

Vieira et al. (2011) reported the superelastic property of laser-welded NiTiNol samples by investigating the mechanical performance with the help of cyclic loading and unloading tests for 30 cycles with strain values ranging up to 8%. They observed better functionality with superior recoverable strain of the welded joints for a strain value of more than 6%. The softening phenomenon at the beginning and at the end of SE plateau was originated from laser welding and was the cause behind the stretched-out shape of the loop. The smooth transition was basically related with stress accommodated with slip in the HAZ area before reaching the conversion plateau.

Chan et al. (2012b) carried out a unique study to show the mechanical behavior of the laser-welded joint under cyclic loading condition through constitutive modeling. The generated model was in good concurrence with the

experimental data for two different constant strain values. The transformation sequence and the behavior indicating the SME were stable excluding the primary cycles for both the cases of simulation and experiment. In the simulation, only the presence of austenite in the welded region was considered, but actually, the presence of martensite was also reported in that zone.

Oliveira et al. (2017) reported the mechanical cyclic loading–unloading test of laser-welded NiTiInol plates for 600 cycles for a strain value up to 10%. It was reported that the samples showed great resistance to fatigue during cyclic loading–unloading test for a large value of strain of about 8%–10%. Although stress plateau indicating the superelastic property was formed at 350 MPa stress value for the first cycle, a very small plateau was also observed at the very starting of the cycle due to R-phase transformation. This was not perfect situation to show the absolute superelastic property of the joint.

Fatigue behavior of laser-welded NiTiInol wire was studied by Chan et al. (2013a) through small strain-rate cyclic bending measurement. They reported the improvement of fatigue life by considerable reduction in surface strain during cyclic bending test, and the fatigue life was reduced by an increase in bending frequency because of a higher amount of generated heat and more amount of stress-induced martensite formation. In the case of laser-welded samples, the fatigue life was decreased considerably in comparison with base material, although a considerable amount of fatigue resistance was present in the welded samples.

Microhardness Property

The microhardness value varies across the weld-bead depending on the thermal cycle, base material characteristics, and thermomechanical property of the material [Frick et al., 2005; Weinert & Petzoldt, 2004]. It was reported that the maximum value of microhardness was reached at the base material, but it was strikingly reduced in the HAZ and the decreasing trend was also followed in the weld-zone. The same kind of variation trends of microhardness was obtained by Chan et al. (2011) and Mirshekari et al. (2013). This variation of microhardness across different regions was caused by the modification of grain structure after recrystallization phenomena in that region. Cooling rate also significantly played a role in the deviation of microhardness as it affected the grain structure during cooling phase. As the base material had the highest rate of cooling in comparison with FZ and HAZ, it had the finer grain structure contributing to high value of microhardness of this region [Zeng et al., 2016 and Maletta et al., 2009].

Corrosion Property

Yan et al. (2007) made comparison between the corrosion resistance behaviors of laser-welded NiTiInol specimens with un-welded one in 0.9% NaCl solutions.

The elevated open-circuit potential, high polarization resistance, high breakdown potential, and wider passive region of welded samples indicated better corrosion resistance behavior in comparison with base metal.

Chan et al. (2011c) showed the potentiodynamic polarization (PDP) curves of both the base material and welded samples. The welded samples showed better corrosion resistance than base material because of the homogenization of microstructure in the weld-zone.

Chan (2017) revealed that HAZ was the weakest in corrosion resistance behavior due to the coarser grain structure. Stress corrosion cracking phenomena in the welded samples took place at a potential very close to the pitting potential of that region, but the base material showed no corrosion at that potential.

Chan et al. (2013a) investigated an in vitro study on the biocompatibility of laser-welded NiTiNol. They analyzed how the behavior of mesenchymal stem cells was affected by the induced effects of laser welding such as surface roughness, Ti-to-Ni ratio, and topography. The laser-welded surface showed good biocompatibility due to good adhesion and spreading of the cells on the surface. A good spreading of cells indicates the non-toxic nature of the medium. Modification was also observed due to the formation of the grown stem cells in FZ as the FZ had dendritic structure.

In contrary to the above discussion, Hsu et al. (2001) investigated considerable decrement of the corrosion behavior of laser-welded NiTiNol samples in artificial saliva. The amount of corrosion of the FZ was found to be 274 mm/year, which was much higher than 17 mm/year for the base material. This reduction might be due to the formation of intermetallic phases during welding. Hence, proper control of laser welding parameters should be carried out in order to evade the generation of undesirable phases in the weld-bead.

Effect of Laser Welding on Some Other Property

Microscopic Analysis

Laser heating and consequent cooling process definitely affect the microstructure of the laser-welded NiTiNol samples. It was reported that fusion boundary had small columnar grains surrounded by recrystallized HAZ grains [Tam et al., 2011]. They also showed that the dendritic grains were formed in the FZ having perpendicular orientation to the cross-section.

Gong et al. (2011) reported that the surface formation was in good shape with uniform ripple. They depicted the four different regions of the cross section of the welded area: (1) the center of the FZ, (2) the boundary or edge nearer to the top layer of the FZ, (3) the interface or edge nearer to the bottom surface of the FZ, and (4) base material, respectively. They witnessed that

the microstructure of the center of the FZ was fine equaxed in nature, but the microstructure of both the edge zone, nearer to the top or bottom surface, was columnar in nature. The upper surface edge zone columnar crystal was having a stronger direction than the lower one. No noteworthy coarse grain was found in the HAZ region. As the energy density of the laser was high enough, the heating and cooling were also very fast. Due to this phenomenon, recrystallized HAZ was observed beside the fusion boundary. The low thermal conductivity of the alloy prevented the temperature gradient to spread much from the weld center. The highest value of it was found close to the fusion interface and helped in the formation of columnar crystals.

According to Chan et al. (2011), the mode of solidification in the fusion region was varied from planar to cellular and finally cellular to dendritic, as the cooling rate was increased from weld center to base material and the rate of solidification was also improved because of the lowest temperature gradient. The cellular grains started growing from parent NiTiInol grains at the interface in the direction [110] during the laser welding process. The interfacial grain structure was determined by base material and welding conditions as the solidification process instantaneously started from the partially melted base material grains.

According to Gugel et al. (2008), the center of the bead was formed with coarser grains having 50 μm dimensions approximately. The epitaxial grain growth in the FZ occurred in the route of the highest temperature gradient. The grain size was decreased significantly in the HAZ and the approximate dimensions of grains in HAZ were 30 μm with equaxed structure.

Phase Analysis

Till date, various research groups performed different characterization techniques to recognize the occurrence of diverse phases in the different areas of the laser-welded samples. Fernanades et al. (2013), Oliveira et al. (2016), and Oliveira et al. (2015) utilized synchrotron X-ray technique in the transmission mode of operation with good spatial resolution and reduced spot diameter in order to investigate the laser-welded parts from the base material. It was clear that an important microstructural change occurred as austenite and martensite phases were coexisted in both the FZ and HAZ, but the base material composition was fully austenitic in nature at room temperature. The same kind of interpretation was given by Iijima et al. (2008) by the use of conventional CuK α X-ray radiation.

The occurrence of the martensite phase in the thermally affected region was higher; whereas the intensity of austenite in the thermally affected zone got reduced. The peaks corresponding to the martensite phase were present on

both sides of the main peak of austenite, i.e., (110) in case of thermally affected region [Oliveira et al., 2016]. The XRD analysis was also carried out for NiTiInol welded samples at a temperature of 150°C over the austenite finish temperature. As a consequence of the high temperature, the martensite phase disappeared from the weld-bead and HAZ, although this temperature was not adequate to eradicate the other precipitates from this region [Oliveira et al., 2016].

The fraction of martensite phases for two different laser-welded NiTiInol samples was studied by Oliveira et al. (2017). The phase fraction was found to increase from base material toward weld center through the HAZ area. This phase fraction was varied from 4% to 20% in the heat-affected region, whereas it increased sharply above 50% in the FZ. The content of martensite was extremely different for the two zones because the favored volatilization phenomena in the FZ could alter the chemical composition of the base NiTiInol. The identical mechanism occurred for both the samples with slight differences in their value because of variations in heat input and the durability time of Ni-rich precipitates in the HAZ of different samples.

The detection of different phases through structural characterization was very important as these phases played a vital responsibility in the mechanical performance of the laser-welded joints. The generation of different Ti-rich brittle intermetallic compounds such as Ti_2Ni in the thermally treated region was the main cause behind the degradation of the mechanical performance of the joint [Tuissi et al., 2004; Schlossmacher et al., 1997]. Schlossmacher et al. (1997) anticipated that both the occurrences of oxygen in the FZ and Ti_2Ni precipitation were the main causes behind the early and drastic failure of the joint. This statement emphasized the requirement of the NiTiInol base material to be free from any contamination and needed good gas shielding environment to avoid any kind of gas (such as O_2 , H_2) pick up by the molten pool. The occurrence of Ti_2Ni was also harmful for the corrosion resistance property of the joints, as the precipitation sites of Ti_2Ni acted as the favored zone to initiate the pitting phenomenon in the NiTiInol alloy [Dutta et al., 1993]. Other Ni-rich intermetallic compounds such as Ni_3Ti [Zhao et al., 2008a, 2008b] and Ni_4Ti_3 [Li et al., 2010; Oliveira et al., 2016] were also existed in the FZ of NiTiInol.

Effects of Shielding Gas

The shielding gas environment affected the depth of penetration precisely. Weld-bead profile obtained for different heat input at two different gas environments, namely argon and helium, was reported by Chan et al. (2011). Argon gas showed a higher depth of penetration than helium for the same input process parameters because of the dissimilarity in thermophysical properties of both the gases.

Fracture Surface Analysis

The fracture surface image of tensile test samples carried out on the laser-welded as well as un-welded NiTiNol sheets revealed that the un-welded samples went through ductile mode of failure. The fracture surface image of the welding done at 0.6 kW power presented a smooth dendritic solidification structure. Although the fractured surface image of the welding carried out at 0.9 kW power revealed a coarser fractured surface, the close-up image of the same joint represented a dimpled structure, suggesting a ductile intergranular mode of failure [Khan et al., 2008].

Ductile mode of failure can take place in the case of cyclic loading–unloading test because of a variety of cycles and strains. The high strength of the joint could be understood due to the creation of stress-induced martensite. A profound parallel secondary crack was generated on the fractured surface, having dimpled structure, of the similar NiTiNol joint obtained by using the uniaxial tensile test [Vieira et al., 2011]. It might be formed due to the uninterrupted deviation of the locally created stress during the fracture mechanism.

NiTiNol superelastic wire was welded with similar tube with the help of Nd:YAG laser by Zeng et al. (2016), and they revealed the occurrence of the fine dimpled structure on the same region when magnified reasonably. No considerable difference was observed in the chemical composition of the base material and FZ through energy dispersive spectroscopy analysis.

Effect of Welding on Phase Transformation Temperature

Laser welding considerably affects the phase conversion temperature of the SMAs. This change can be calculated by electrical resistivity [Uchil et al., 2002; Antonucci et al., 2007], dilatometry [Uchil et al., 2001; Qiu et al., 2009], and differential scanning calorimetry (DSC) analysis [Bradley et al., 1996; Shaw, 2008; Chu et al., 2005].

The changes in phase transformation temperature of NiTiNol wire stimulated by laser welding process could be identified by DSC graphs. From DSC graphs, the phase transformation temperature of both the base and welded samples could be identified by the occurrence of peak in the heating and cooling cycle. The conversion temperatures were generated by the use of the software from the DSC curves. Chan et al. (2013b) reported that both the parent sample and the laser-welded samples were in the austenite phase in room temperature and were capable of showing their superelastic properties. Lowering of the phase transformation temperature occurred for laser-welded samples although phase transformations were observed for both the parent and the welded samples.

Microstructures, grain size, and chemical composition of the FZ were affected by the resolidification process of laser welding. Depending on the Ni-to-Ti ratio in the molten pool, the phase transformation temperature could be changed because of the precipitation of different Ni-rich or Ti-rich intermetallic phases modified the overall distribution of the elements in the FZ. Furthermore, the occurrence of nucleation sites was also dependent on grain size. With boost in grain size, the number of nucleation sites decreased and would also affect the phase transformation temperature. In short, the laser welding process caused the lattice dislocations by the deposition of a variety of intermetallic phases, which basically affected the microstructure and would also escalate the residual stress after welding [Chan et al., 2015; Chan et al., 2011; Frenzel et al., 2010].

Gong et al. (2011) showed the same kind of variation trend of phase transformation temperature of the laser-welded NiTiInol sheet by DSC measurement. Although in the heating cycle of the base material, one stage conversion of the martensite phase to austenite took place, the occurrence of R-phase around 43.81°C was obvious in the cooling cycle, which designated to the austenite to martensite transformation. The final transformation of R-phase to martensite occurred around 0°C. In case of laser-welded samples, the existence of R-phase was prominent but the location of the peak associated with R-phase was changed due to the precipitations of different intermetallic phases in the welded region [Zhou et al., 2005]. Contrary to this, the phase transformation behavior of the annealed laser-welded NiTiInol sheets did not reveal many differences compared with the parent reference material without showing noteworthy loss of SME [Tirry & Schryvers, 2005; Schryvers et al., 2006].

As the laser welding process was significantly affected by different process parameters, the phase transformation temperature could also be affected by different operating parameters mainly power and scan speed. Zamani et al. (2017) showed how laser power affected the phase conversion temperature of the laser-welded NiTiInol wire. It was observed that with an increase in laser power, the phase change temperature also increased. This happened due to the direct effect of laser power on the rate of evaporation of NiTi and Ni. The selective evaporation of any of the constituents could substantially affect the functional property of the alloy. The SME of the NiTiInol was very much dependent on the constituents of the alloy. A minor modification in the composition may affect the functional properties greatly.

Effect of PWHT on Laser Welding of NiTiInol

Post-weld heat treatments (PWHT) were capable of improving the tensile properties as well as reducing drawbacks associated with the generation of

residual stresses during laser welding of the NiTiInol alloy [Oliveira et al., 2017; Cho et al., 2004]. Therefore, selection of heat treatment parameters such as temperature and time for heat treatment played an essential role in the recovery of the mechanical properties. It was proven that PWHT could be useful to laser-welded NiTiInol joints in a positive way to eliminate the bad impact of laser welding [Fernandes et al., 2017]. The main disadvantage of laser welding of NiTiInol was the noteworthy drop of SME and the superelastic property of the FZ. This was caused mainly by the contamination and isolation of the solute in the solidification phase or because of the presence of microstructural inhomogeneity in the different thermal-treated regions and the base material [Otsuka & Ren, 1999]. Depending on available time and aging temperature, different kinds of Ni-rich intermetallic phases such as Ni_3Ti , Ni_4Ti_3 , and Ni_3Ti_2 might emerge in the FZ [Nishida et al., 1986; Paula et al., 2008]. Among all the phases, Ni_4Ti_3 , the most considerable secondary phase, emerged at a low level of temperature and shorter time duration and considerably affected the SME and SE of the NiTiInol alloy [KE et al., 2012]. In the next part, the in-depth discussion of the effect of PWHT on different laser welding properties would be considered.

Effect on Microhardness

Precipitation hardening phenomena and grain dimension largely control the microhardness value. With increase in PWHT, the grain size became larger but the microhardness value decreased by the Hall–Petch effect. Additionally, the precipitation of Ni_4Ti_3 was increased by the higher value of PWHT temperature. This precipitation of the secondary phases augmented the effect of precipitation hardening phenomena. It was obvious that the microhardness value of the base material was more than that of the FZ for the same PWHT parameter setting. The maximum microhardness value was obtained for heat treatment done at 623 K for both base material and FZ. From the figure, it was evident that though the base material was harder than the welded sample, the PWHT done at 623 K enhanced the microhardness of the weld-bead and the maximum value was in the range of 340 HVN_{0.05} [Chan et al., 2012c].

Effect on Tensile Property and Fractured Surface

Yan and Ge (2014) showed the result of annealing on the tensile property of the laser-welded NiTiInol wires. Although the nature of all the curves was similar and experienced stress-induced martensite formation, it was found that with the rise in annealing temperature, the plateau stress showed a minor variation, more significantly a decreasing trend. However, the maximum tensile strength of the as-welded samples was about 800 MPa, and a trivial improvement in the

maximum tensile strength was observed at 400°C. On the other hand, it was found that with higher value of annealing temperature, the maximum tensile strength value was reduced. A similar observation was reported by Chan et al. (2012c). This reduction could be recognized by the drop of the tensile property induced by the defects originated from thermal treatments and the change in micro-structure and grain size of the FZ. As a result of all these, the zone was acted as the favored zone for the generation of fracture under tensile loading condition [Chan et al., 2012c].

Yan et al. (2007) showed that though there was not much difference in the mode of fracture for two different heat treatment schedules, the PWHT at 400°C produced a much smoother surface than the others. All the samples showed brittle mode of fracture as river like patterns were prominent on the fractured surface.

Effect on Cyclic Deformation Behavior

Panton et al. (2016) reported that though the base material could withstand the maximum number of thermomechanical cycles with maximum actuation, the welded samples showed a razor-sharp drop in the actuation mechanism with less value of cycle failure and elevated necking phenomena. On the other hand, PWHT positively affected to enhance the actuation and cycle life of the deteriorated welded samples. The built-up dislocation phenomena helped to reduce the transformation strain. Even the dislocation phenomena played important role in the plastic deformation of NiTiNol and the necking phenomena.

The effect of diverse PWHT methods on the superelastic property of the laser-welded NiTiNol samples was reported by Yan and Ge (2014). The cyclic deformation behavior was analyzed for the laser as-welded samples and post-welded heat treatment samples carried out at temperatures of 400°C, 450°C, and 500°C for 1 h time. The residual strains for all specimens were enhanced by increasing the number of cycles. The samples with PWHT, done at 400°C and 450°C, showed a minimum amount of residual strain in comparison with the as-welded samples. On the other hand, the PWHT done at 500 °C revealed extra amount of residual strain with added number of cycles among all the specimens. It was also clear that the stress of all specimens were decreased by increasing the cycle number.

Yan et al. (2007) also demonstrated how thermal treatment affected the fatigue behavior of laser-welded NiTiNol wire. It was shown that the PWHT at 400°C resulted in improved fatigue resistance property than that of PWHT carried out at 500°C in rotating-bending fatigue test. The improvement in fatigue resistance property was due to the difference in dimension of Ni_4Ti_3 precipitates in the PWHT area compared with the untreated zone. Actually, the

precipitation of smaller dimension particles (about 10 nm) helped to recover the fatigue resistance property of the alloy. Ultimately, the precipitation of the secondary phase of Ni_4Ti_3 reduced the fatigue resistance property by making the area coarser.

Chan et al. (2012) also stated the effect of PWHT on the cycling behavior of laser-welded NiTiInol joints. Although the PWHT performed at 350°C showed the presence of superelastic property, the same PWHT carried out at 450°C only identified the martensitic nature of the curve. The non-existence of the stress crack was also reported. From this, it was clear that PWHT introduced some homogeneous phases to the present phases of the FZ. The observed irrecoverable strain was caused by the martensitic detwinning phenomena and the stability of that particular phase at the test temperature along with austenite phase. The PWHT introduced an enhancement to the joint in respect of irrecoverable strain when compared with non-heat-treated samples. This could be the cause behind the fine grain structure of Ni_4Ti_3 after the heat treatment process. The irrecoverable strain was more prominent in case of PWHT done at 450 °C because of the changes induced in the phase conversion temperature permitting the stability of martensite at room temperature along with some amount of austenite phase. An appropriate PWHT parameter setting was essential for the conservation of functional aspects of the welded joints.

Effect on Corrosion Resistance Property

Chan et al. (2012c) reported the heat treatment effect on the pitting corrosion property of the laser-welded NiTiInol wire. This study, concentrated on the influence of heat treatment, was analyzed by the comparison of the corrosion resistance property between the base material and laser-welded samples in Hank's solution at normal body temperature, i.e., at 37.5°C. It was witnessed that the PWHT recovered the corrosion resistance property as the heat treatment process could modify the elemental constituents of the surface of the welded parts when the temperature of the surface treatment was 350°C or more than that. It was also identified that the PWHT at 350°C affected more the precipitation of the secondary intermetallic phase, namely Ni_4Ti_3 in both the weld-bead and base material, than the PWHT done at 450°C. Even the PWHT helped to considerably increase the Ti-to-Ni ratio and the TiO_2 -to-TiO ratio on the surface layer, which in turn assisted to develop the corrosion resistance property of the welding [Chan et al., 2011; Chan et al., 2012c].

Chan and Man (2013) compared the cyclic behavior of NiTiInol base material and laser as-welded samples, heat-treated samples at 350°C for 60 min, and up to 4% strain for ten cycles in slow strain-rate test in Hank's solution.

Enhanced corrosion resistance behavior of the heat-treated samples was guaranteed by the upper value of the OCP, and the stable pattern of the same for heat-treated samples than the as-welded samples. Nevertheless, the corrosion property of the thermally treated samples was lesser than that of the reference base sample in Hank's solution. The heat treatment process helped to decrease the accrued irrecoverable strain in a significant way when compared with as-welded samples in slow-rate cyclic test in Hank's solution. In this way, better superelastic property could be achieved. Sufficient PWHT helped to achieve lower residual stress, when the welded joints were gone through stress corrosion conditions and the effect was beneficial for the weld performance.

Effect on Microstructure

The effect of PWHT on the microstructure of the welded zone and the HAZ was established through the micrographs by Chan et al. (2012c). In the weld-bead though the presence of primary dendritic structure was prominent, the presence of second and third arms was exceeded. This was caused more in the transverse direction than the longitudinal grain development direction. The equiaxed grain structure was visible in the HAZ. Instead of pure austenitic structure, a twinned martensite structure was observed in the FZ as well as in HAZ. This was how the microstructure of the welded NiTiNol could be affected by proper selection of PWHT temperature.

Effect on Phase Transformation Temperature

The phase transformation temperatures of the parent material, as-welded samples, and PWHT samples of NiTiNol are presented by Panton et al. (2016), based on the acquired results of differential scanning calorimetry analysis. The peak associated with the austenite phase of the base material was basically the combination of martensite to R-phase and R-phase to austenite phase transformation. The generated R-phase basically experienced thermomechanical treatment to augment nanostructures of precipitates and grains and also to enhance the density of dislocations [Elhania et al., 2012]. These shortcomings produced a comparatively lower strain value for the intermediary R-phase conversion than the direct austenite to martensite phase conversion [Panton et al., 2016; Waitz et al., 2008]. It was observed that the PWHT, related with the entire gauge length, had negligible effect on the grain growth phenomena of FZ and HAZ. However, the base material in gauge length directed to reduce the density of the dislocations and thus reduced the UTS. The ductility of the PWHT samples was enhanced in comparison with the as-welded samples and base material. In fact, the R-phase plateau was formed by the modification of the phase transformation temperature, and the PWHT had trivial effect on

the development of the detwinning phenomena of martensite and the early R-phase plateau [Chan et al., 2012 (c); Panton et al., 2016].

In case of Ni-rich alloy, PWHT endorsed the phase transformation temperatures because of the reduction of Ni in the matrix of the material neighboring the secondary Ni-rich phase [Otsuka & Ren, 2005]. This was also experimentally shown by Chan et al. (2012c). The PWHT improved the phase transformation temperature of the FZ by the elimination of different defects associated with dislocation phenomena and reduction of Ni. On the other hand, the formation of different brittle intermetallic precipitations in Ti-rich welded samples could encourage a decrement in the phase conversion temperatures. It was clear that the R-phase was the intermediate phase during the conversion sequence of the main two phases. Alike explanations were published by Yan et al. (2014) and the R-phase transformation was also validated by the occurrence of Ni_4Ti_3 precipitations that helped to endorse the survival of the coherent stress fields supporting the development of the phase [Sitepu et al., 2002].

Summary

NiTinol has some weldability issues related to grain size difference across different zones of the joint, precipitation of brittle intermetallic compounds, and variations of composition in thermally treated regions due to favored volatilization of some of the constituents.

- Due to the superior aspects of laser welding process, a large number of studies were concentrated on the different features of laser welding of NiTiInol. A detailed discussion was dedicated to different aspects of laser welding of NiTiInol including microstructural changes, metallurgical changes, mechanical properties, functional properties, and corrosion resistance properties.
- Moreover, different welding parameters, most significantly scan speed and laser power in the case of laser welding, have strong effects on mechanical properties, microstructural properties, functional properties, and corrosion resistance properties of the NiTiInol-welded joints. Therefore, attaining the optimum welding parameters would help to eliminate the probable disadvantages incorporated due to laser welding process.
- Though different aspects of laser welding of NiTiInol were studied by different researchers, the effects of different laser parameters in phase formation and intermetallic formation and their correlation with different thermomechanical properties are not reported in detail in existing

literature. Study on the thermal cycles and the correlation between the thermal signal and the formation of different intermetallic compounds and their correlation with different thermophysical properties could be one of the future scope of laser welding of NiTiInol.

- The other future scope includes the evaluation of residual stress in laser-welded NiTiInol because of the diverse microstructure within and surrounding the weld and the rapid thermokinetics associated with the laser-based processes. The correlation between the residual stress and the formation of stress-induced martensite formation also includes the future scope of the study.

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Chapter 5

Influence of Process Parameters on Weld Quality and Evolution of Microstructure and Microhardness in Laser Welding of NiTiInol-SS304 Dissimilar Combination

**S. Datta¹, M. S. Raza¹, K. Chirikuri¹,
P. Saha¹ and D. K. Pratihari¹**

¹Department of Mechanical Engineering, Indian Institute of Technology Kharagpur,
Kharagpur – 721302, India

Abstract

Both NiTiInol and SS304 are extensively used in biomedical industries. Dissimilar welding of NiTiInol with SS304 is not widely reported. In this experimental work, dissimilar combination of NiTiInol and SS304 for joining with the help of fiber laser was studied. The purpose of this experimental work was to recognize the influence of laser welding process variables on the quality characteristics of dissimilar joint of SS304 and NiTiInol. Laser power and scan speed were chosen as input process parameters, and top bead width, bottom bead width, and microhardness of the weld area were considered as output variables. Statistical regression investigation was performed to determine the input–output relationship. The obtained equation was found to be adequate for predicting the output variables within the working range of the input variables.

Keywords: Laser welding; Dissimilar welding; NiTiInol; SS304; regression analysis

Introduction

The equiatomic Nickel–Titanium shape memory alloy (NiTiNol) plays a vital role in biomedical engineering industry, because of their superior characteristics like pseudoelasticity, shape memory properties, corrosion resistance performance, and biocompatibility [Duerig et al. 1999; Hall 2005; Pelton et al. 2003]. It is widely utilized in the manufacturing of medical instruments for invasive surgical treatment, such as stents, catheter, guide wires, and orthodontic arch wires. To increase their choice of use, there is a robust possibility of joining NiTiNol alloys with some other alloys, such as platinum, titanium, and stainless steel (SS), having biocompatibility. Stainless steel is also considered as a good biomedical alloy for its superior mechanical characteristics, low cost, workability, and good corrosion resistance [Uenishi et al. 2003]. The stainless steel and NiTiNol combination would be desirable in biomedical industry for hybrid mechanical properties and low cost. In case of NiTiNol and SS composite orthodontic arch wires, the advantages from both the constituents are considered. The NiTiNol part delivers suitable orthodontic force to teeth, and from SS portion adequate anchorage for supporting the teeth can be obtained. This combination significantly increases the effectiveness of the orthodontic therapy and is helpful in relieving the pain of patients. The huge difference in chemical and thermophysical characteristics among these two materials creates problem in different material welding and makes the process complex and difficult. As Ti is highly reactive, the tendency of forming brittle intermetallic compounds with Fe, Cr, and Ni, when NiTiNol is welded with SS, is very high. Brittle intermetallic compounds, such as TiCr_2 , TiFe_2 , and TiNi_3 , are generated in the weld pool and are the cause behind the low tensile properties and the brittle failure of the joins [Qiu et al. 2005; Li et al. 2012; Li et al. 2009; Vander Eijk et al. 2003; Pouquet et al. 2012; Hall 2004].

Poor machinability, formability, and available joining techniques of NiTiNol in similar and dissimilar material combination restrict its widespread application in different fields of engineering. So far different joining methods including resistance spot welding, friction welding, tungsten inert gas welding, etc., were widely reported for similar and dissimilar welding of the NiTiNol alloy. It was found that laser welding, a significant manufacturing process, has several benefits as a joining technique of Nitinol over other extensively applied joining methods. The process can be characterized by narrow weld width and heat-affected zone (HAZ), good bead quality, and high depth of penetration. Great power density value makes the process an important joining technique [Steen & Mazumder 2010]. The input process parameters of laser welding govern the shape of weld-bead, and the heat input of the process is controlled

by input parameters [Dawes 1992]. For a good quality weld, the input power, shielding gas supply, focal position, welding speed, and position accurateness should be appropriately chosen [Long et al. 2016]. Response surface methodology (RSM) is an extensively utilized technique to envisage the weld-bead geometry and mechanical characteristics in several welding processes [Kim et al. 2002; Gunaraj et al. 1999; Benyounis et al. 2004; Gunaraj et al. 2000; Allen et al. 2002; Benyounis et al. 2008; Acharjee et al. 2009; Wang et al. 2012; Zhao et al. 2012; Benyounis et al. 2005; Kanigalpula et al. 2016]. As RSM can generate quadratic relation between input and output process variables, can do optimization of the process parameters with minimum number of experimental trial, and can investigate the interaction effect between different process variables, this method is widely utilized in different engineering experimentation. The complex correlation between controllable input process parameters and the response of the welding problem could be easily obtained using the empirical method of RSM technique.

Both NiTiInol and steel have good biocompatibility and are widely used in the field of biomedical application. Successful joining of these two materials can reduce the overall cost of many biomedical implant devices. Because of the difference in thermophysical and mechanical properties of both the materials, joining of these two materials is not very easy. The composition of the weld-bead is basically controlled by the amount of melting of both these materials. This basically controls the overall joint strength of dissimilar welding by controlling the formation of different phases and the solidification structure.

In this experimental work, the dissimilar joint of SS304 and NiTiInol sheets of 1 mm thickness was performed using a Yb-fiber laser system. The profile geometry of weld-bead, microstructure, microhardness, and possible creation of new phases during the laser welding was investigated. Furthermore, RSM was utilized to make models to predict weld-bead profile (top width, bottom width) and microhardness of the bead for butt-welding of the SS304 and NiTiInol by continuous wave (CW) fiber laser.

Materials and Methods

NiTiInol (52 atomic% Ni and 48 atomic% Ti) and SS304 (9 atomic% Ni, 19 atomic% Cr, 1.8 atomic% Mn, 0.23 atomic% Si, 0.03 atomic% S, 0.08 atomic% C, and balance amount is Fe) of 1 mm thickness with sample size of 50 mm× 25 mm were utilized as workpiece for laser beam welding in butt-joint configuration. Trial experiments of butt welding were executed by altering one process variable at a time to regulate the employed limit of each variable. Full

depth of penetration was the condition for choosing the working range. The experimentations were performed using a 2 kW fiber laser system following the design matrix in a random manner to evade any orderly inaccuracy. The laser was operated in continuous mode. The laser beam was focused on the top surface of the workpiece. Argon gas supply was employed as shielding system at a flow rate of 7 lpm to safeguard the molten material from atmospheric contaminations.

The experimentation was performed depending on a three-level central composite design (CCD) with three replications. Laser power (800 W–1200 W) and welding speed (3500–4500 mm/min) were taken as self-regulating input process variables. Table 5.1 displays the input process parameters and investigational design levels utilized for the experimentation. RSM was used as the design of experiment tool by statistical software, Minitab 15. Table 5.2 shows the diverse arrangement of process parameters used in this experimentation. Table 5.3 illustrates the dissimilar arrangement of process variables utilized for testing the competence of the used mathematical model. Statistical regression analysis was conducted to evaluate the input–output correlation. The measured output parameters were top-width (TW), bottom-width (BW), and microhardness (μH) of the bead. Statistical significance of every term in regression equation was scrutinized using different adequacy measures (such as sequential F test, lack-of-fit test) with the help of the same software. The statistical regression equation for input–output relation is represented by:

$$Y = \beta_0 + \sum_{i=1}^k \beta_i X_i + \sum_{i=1}^k \beta_{ii} X_i^2 + \sum_{i=1}^k \sum_{i < j}^k \beta_{ij} X_i X_j + \epsilon \tag{1}$$

where X is the independent input variables; Y is the dependent response variables; $\beta_0, \beta_1, \beta_2, \dots$; and β_{ii} are the factors to be calculated by means of a least square method. ϵ represents the error in the fitting.

Two crosswise samples were cut from each weldment. Typical metallographic samples were prepared from each crosswise section. The bead profile geometry parameters were calculated using zoom optical microscope (Make: Zeiss). Microhardness variations across the bead were measured using

Table 5.1 Process parameters and their ranges used for experimentation

Input	Coded symbol	Uncoded symbol	Min. value	Mid. value	Max. value
Laser power (W)	X1	P	800	1000	1200
Scan speed (mm/min)	X2	v	3500	4000	4500

Table 5.2 CCD table for laser beam butt welding of NiTiInol and SS304

Sl. No.	P (W)	v (mm/min)	TBW (μm)	Standard Deviation TBW	BBW (μm)	Standard Deviation BBW	μH (HV _{0.05})	Standard Deviation μH
1	1000	4000	1060	9.73	642	2.43	851.3	6.71
2	1000	4000	1075	5.65	662	4.29	841.68	8.89
3	1000	4000	1089	3.32	634	5.32	853	7.68
4	1000	4000	1074	6.79	625	6.38	863	10.32
5	1000	4000	1091	8.58	617	5.28	855	4.87
6	800	4000	960	4.65	513	4.38	820	6.48
7	800	4500	819	7.73	423	8.78	780	9.49
8	1000	3500	1205	2.96	794	10.68	875	7.54
9	800	3500	1175	10.21	623	9.27	860	8.36
10	1000	4500	928	8.23	568	7.27	749	3.38
11	1200	3500	1328	5.76	968	10.98	892	9.23
12	1200	4500	1156	6.65	752	7.86	836	4.7
13	1200	4000	1247	4.32	876	3.85	879	5.69

Table 5.3 Test conditions for several arrangements of input process variables

Sl. No.	P (W)	v (mm/min)	TBW (μm)	Standard deviation TBW	BBW (μm)	Standard deviation BBW	μH (HV _{0.05})	Standard deviation μH
1	900	3750	1100	5.82	636	9.25	850	4.54
2	900	4000	1033	4.95	582	7.63	829	9.77
3	900	4250	966	9.38	529	4.53	808	7.57
4	1100	3750	1225	10.74	808	6.43	877	5.78
5	1100	4000	1158	8.36	755	2.78	856	8.49
6	1100	4250	1092	7.39	701	8.73	835	9.65

a Vickers's microhardness tester [Model: S. Auto, Make: Omnitech] following the ASTM E384 standard with a load of 50 g and a dwell time of 10 s. The average of the microhardness value inside the bead was considered as the microhardness value in the modeling and optimization study. Scanning electron microscopy (SEM) [Model: EVO 18 Research, Make: Zeiss] was utilized to disclose the microstructures. X-Ray diffraction (XRD) analysis, using Cu-K α radiation in the PANalytic EMPYREAN system, of welded and the parent materials was done to acquire knowledge about the creation of different new phases in the course of welding.

Results and Discussion

General discussion, metallurgical characteristics of fusion zone, and microhardness variation

Full depth of penetration was confirmed by visual inspection of each sample of each experimental trial. The overall strength of the welding was not adequate, and in some cases, the parts were disjointed in the course of unclamping. The swirling effect generated by the circulation of the molten material inside the weld-bead was evident in each and every welding (refer Figure 5.1). Because of the interaction of the laser with the workpiece, quick heating, melting, and recirculation of the liquefied material have occurred. The swirling of the molten material inside the bead was triggered by the stirring action of different forces and the rapid heating and cooling characteristics of the process. The heat transfer rate and the flow of the liquid material decided the

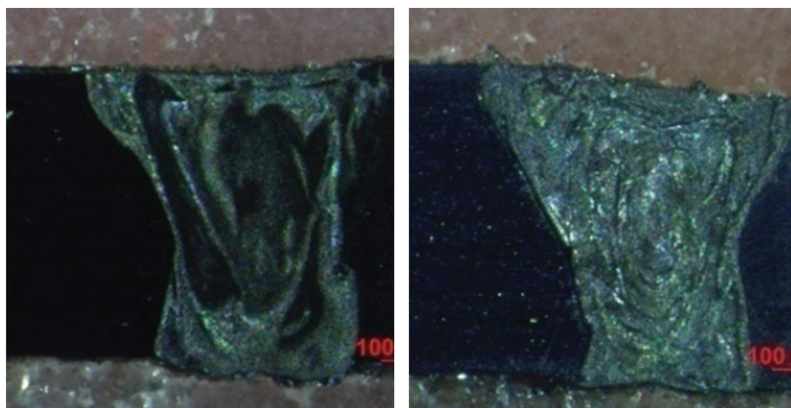


Figure 5.1 Weld-bead geometry of different beads presenting the swirling effect.

overall geometry of the weld, temperature slope, local cooling rate, concentration gradient, microstructure, composition, and the overall characteristics of the weld. The weld-bead shape undoubtedly specified the conduction mode of welding. The top surface of the SEM image of every weld-bead showed the presence of shrinkage cavity, pores, and transverse cracks during welding (refer Figure 5.2). The presence of microscopic cracks, near the NiTi-portion of the welding, was found in most of the welding performed at high power and scan speed after the completion of complete cooling (refer Figure.5.3).

From XRD analysis, it was found that different brittle intermetallic phases of Ni, Ti, Mn, and Fe were largely existent in the whole weld-bead. Because of the widespread existence of these brittle intermetallic phases, the overall strength of the weld was considerably poor. To enhance the strength of the weld, the generation of such intermetallic phases should be stopped. Figure 5.4 is basically showing the XRD pattern of the weld-bead.

Figure 5.5 shows the microhardness variations across the weld-bead. It was found that the microhardness value of the weld-bead was considerably higher than that of the base sample. This high microhardness value was indicative of the brittle nature of the weld-bead material. The presence of different brittle intermetallic compounds, such as FeTi and Mn_2Ti_4O , was ascertained by XRD analysis. This could be the reason for the high microhardness value of the weld-bead. The typical microhardness value of the bead was observed to be nearby $750\text{ HVN}_{0.05}$ with the highest value of $900\text{ HVN}_{0.05}$, in the shadowy area of the weld-bead owing to the occurrence of high quantity of intermetallic compounds.

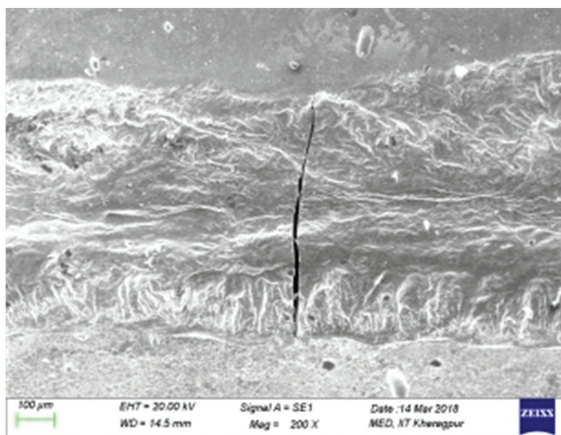


Figure 5.2 SEM image of top-surface presenting the occurrence of crosswise cracks and porosity.

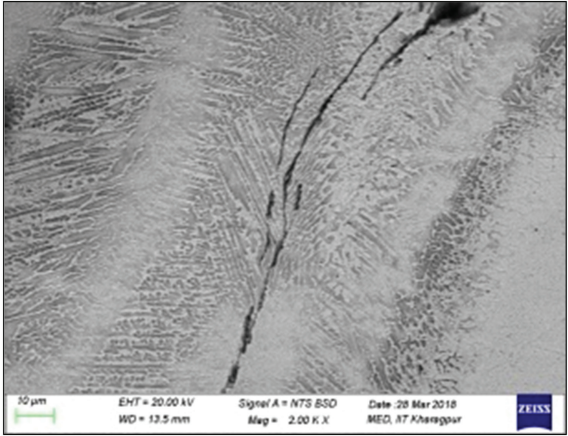


Figure 5.3 SEM image of cross-section of bead, displaying the existence of microscopic cracks, parallel to NiTi side.

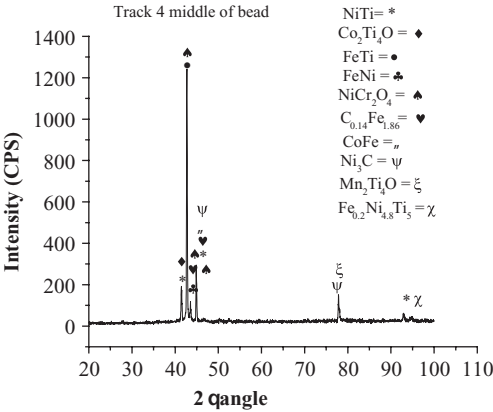


Figure 5.4 XRD patterns of welded samples for sample 7

Input–Output Mappings Using Regression Analysis

Input and output process variables were associated by means of the regression analysis mapping technique. The significance check was conducted to assess the competence of the developed model at a significance level of 95%.

Top Bead-Width (TBW)

Initially, the top bead width was modeled as a nonlinear function of input process variables, which includes laser power (P) and scan speed (v), denoted by X1 and X2, respectively, in the coded form. The significance test results are

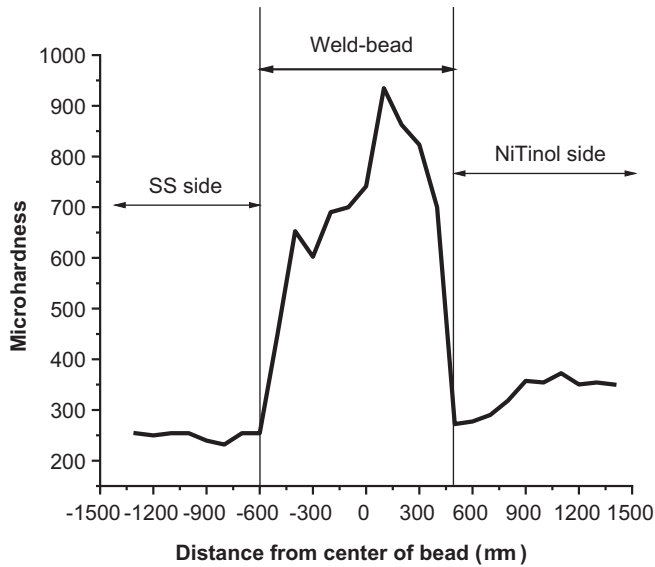


Figure 5.5 Deviation of microhardness across the bead for sample 12.

Table 5.4 Results of significance test for TBW

Sl. No.	Term	Coef.	SE Coef.	<i>T</i>	<i>P</i> value
1	Constant	1075.62	4.932	218.084	0.000
2	X1	125.67	4.849	25.915	0.000
3	X2	−134.17	4.849	−27.668	0.000
4	X1 ²	44.83	7.147	6.272	0.000
5	X2 ²	−3.67	7.147	−0.514	0.623
6	X1X2	46.00	5.939	7.745	0.000

S = 11.8781, *R*-Sq = 99.55%, *R*-Sq (adj.) = 99.22%.

given in Table 5.4. The terms X1, X2, X1², and X1X2 were significant terms. It implied that the relation between TBW and laser power (X1) was nonlinear, but the relation between TBW and scan speed (X2) was linear. The developed model was statistically sufficient for prediction, as the *R*-Sq value (correlation coefficient) was obtained to be 99.55%. The lack of fit was observed to be unimportant from the ANOVA table (Table 5.7). Hence, inconsequential terms might be omitted from the developed model. As the *F* values were in excess of those presented in the typical statistical table, the model was statistically sufficient.

Figure 5.6 demonstrates the contour plot of TBW with P and v . The TBW increased with increasing power and decreased with increasing scan speed. The un-coded form of the equation for TBW was as follows:

$$\text{TBW}_{\text{un-coded}} = 4246 - 3.45305X_1 - 0.610816X_2 + 0.00112069 X_1^2 - 146897 E^{-05} X_2^2 + 0.00046 X_1 X_2 \quad (2)$$

Bottom Bead-Width (BBW)

In case of BBW, the terms X_1 , X_2 , and X_1^2 were observed to be substantial (Table 5.5). It means that the relation between power and bottom bead width is non-linear, but with scan speed the relation is linear. The model was observed

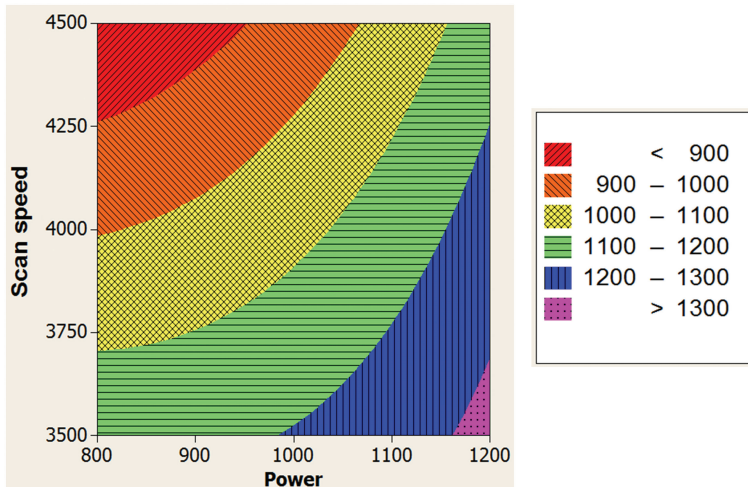


Figure 5.6 Contour plot for top bead-width (TBW).

Table 5.5 Results of significance test for BBW

Sl. No.	Term	Coef.	SE Coef.	T	P value
1	Constant	642.621	8.767	73.303	0.000
2	X_1	172.833	8.619	20.052	0.000
3	X_2	-107.00	8.619	12.414	0.000
4	X_1^2	35.328	12.704	2.781	0.027
5	X_2^2	21.828	12.704	1.718	0.129
6	X_1X_2	-4.000	10.556	0.379	0.716

$S = 21.1130$, $R\text{-Sq} = 98.79\%$, $R\text{-Sq (adj.)} = 97.93\%$.

to be statistically sufficient for prediction, as the R-Sq value (correlation coefficient) was obtained to be 98.79%. From the ANOVA table, it is clear that the lack of fit is irrelevant, so irrelevant terms could be omitted from the model. From the contour plot of BBW (Figure 5.7), it is prominent that BBW increases with rise in P and decreases with increase in v . The un-coded form of equation for TBW was as follows:

$$\begin{aligned} \text{BBW}_{\text{un-coded}} = & 2754.61 - 0.7442213X_1 - 0.872483X_2 \\ & + 0.00088319 X_1^2 + 8.73103 E^{-05} X_2^2 \\ & - 4.000 E^{-05} X_1X_2 \end{aligned} \quad (3)$$

Microhardness of the Bead (μH)

In this case of μH , the terms X_1 , X_2 , and X_2^2 were observed to be important (Table 5.6). It means that the relation between power and microhardness is linear, but with scan speed the relation is nonlinear. The model was found to be statistically acceptable for prediction, as the R-Sq value (correlation coefficient) obtained was 97.14%. From the ANOVA table (Table 5.7), it is clear that the lack of fit is irrelevant, so unimportant terms could be omitted from the model. From the contour plot of μH (Figure 5.8), it is prominent that μH increases with an increase in P and decreases with an increase in v . The un-coded form of equation for μH was expressed as follows:

$$\begin{aligned} \mu H_{\text{un-coded}} = & 333.694 - 0.440589 X_1 + 0.417449 X_2 + 7.94569 E^{-05} X_1^2 \\ & - 7.56569 E^{-05} X_2^2 + 0.000104775 X_1X_2 \end{aligned}$$

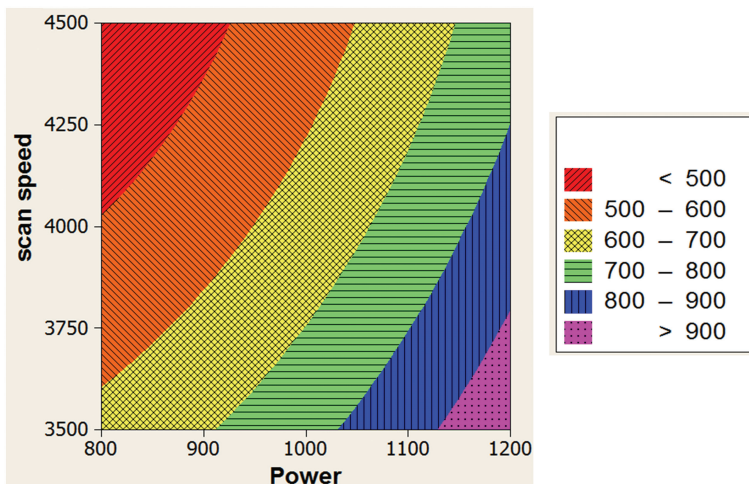


Figure 5.7 Contour plot for bottom bead-width (BBW).

Table 5.6 Results of significance test for μH

Sl. No.	Term	Coef.	SE Coef.	<i>T</i>	<i>P</i> value
1	Constant	850.946	3.445	247.015	0.000
2	X1	27.485	3.387	8.115	0.000
3	X2	-41.516	3.387	12.257	0.000
4	X1 ²	3.178	4.992	0.637	0.545
5	X2 ²	-18.914	4.992	3.789	0.007
6	X1X2	10.478	4.148	2.526	0.039

$S = 8.29648$, $R\text{-Sq} = 97.14\%$, $R\text{-Sq (adj.)} = 95.09\%$.

Test Experiment

Six test cases (given in Table 5.3) were utilized to determine the effectiveness of the established model. Figure 5.9 demonstrates the values of percent deviation in estimating all three output variables. The absolute minimum values of deviation for TBW, BBW, and μH are 1.57%, 3.57%, and 6.62%, respectively. The maximum value for the same for TBW, BBW, and μH are 11.87%, 9.49%, and 15%, consecutively. The typical values of percent deviation in prediction of TBW, BBW, and μH are 6.331, 6.57, and 9.97, respectively. It was found that the developed model was capable to predict the output variables and in line with previous published work [Kanigalpula et al. 2016].

Conclusions

In this experimental work, laser beam butt welding of NiTiInol and SS304 was performed using a fiber laser system. Experiments were carried out following the central composite design, and statistical regression analysis was done to connect the input–output parameters. Some primary metallurgical observations were also accompanied to get a basic idea of physical and metallurgical changes occurred in the course of welding. The succeeding conclusions could be drawn from this experimental observations:

1. From phase analysis, it was prominent that hard and brittle intermetallic phases such as CoFe, FeTi, FeNi, Ni₃C, etc., were generated in the course of laser beam butt welding of these dissimilar material combinations.
2. From the visual inspection, it was found that the obtained joint strength was not enough because of the intermetallic phases and the solidification cracking phenomena that took place after cooling.

Table 5.7 Analysis of variance (ANOVA) for the responses

Source	DF	Seq. SS	Adj. SS	Adj. MS	<i>F</i>	<i>P</i>
ANOVA for TBW						
Regression	5	217351	217351	43470	308.10	0.000
Linear	2	202757	202757	101378	718.54	0.000
Square	2	6131	6131	3065	21.73	0.001
Interaction	1	8464	8464	8464	59.99	0.000
Res. error	7	988	988	141		
Lack-of-fit	3	349	349	116	0.73	0.586
Pure error	4	639	639	160		
Total	12	218339				
ANOVA for BBW						
Regression	5	255456	255456	51091	114.62	0.000
Linear	2	247922	247922	123961	278.09	0.000
Square	2	7470	7470	3735	8.38	0.014
Interaction	1	64	64	64	0.14	0.716
Res. error	7	3120	3120	446		
Lack-of-fit	3	1922	1922	641	2.14	0.238
Pure error	4	1198	1198	300		
Total	12	258576				
ANOVA for μH						
Regression	5	16353.5	16353.5	3270.70	47.52	0.000
Linear	2	14873.9	14873.9	7436.97	108.05	0.000
Square	2	1040.5	1040.5	520.23	7.56	0.018
Interaction	1	439.1	439.1	439.11	6.38	0.039
Res. error	7	481.8	481.8	68.83		
Lack-of-fit	3	247.0	247.0	82.33	1.40	0.365
Pure error	4	234.8	234.8	58.71		
Total	12	16835.3				

- Microhardness of the bead increases considerably, in some cases four times than that of the base material, after welding which may make the joint brittle.
- From the developed model, it is prominent that welding performed with low power and high scan speed generates minimum bead width, and in that case the increase of microhardness to a very high value is somehow restricted. Whenever the microhardness value would not be very high, the brittleness of the joint would be reduced.

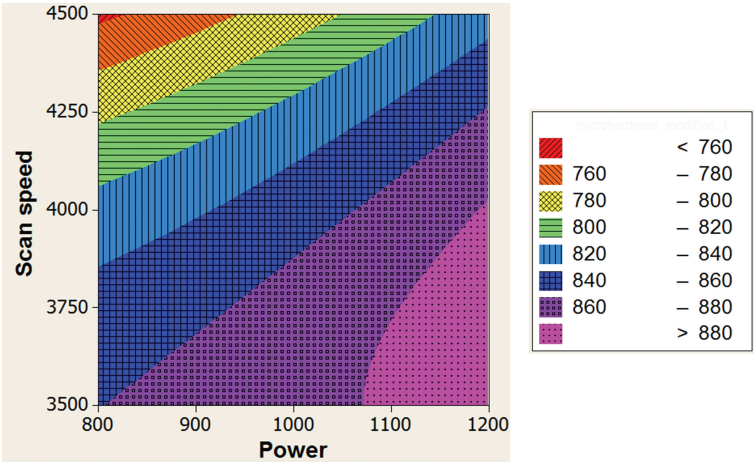


Figure 5.8 Contour plot for microhardness (μH).

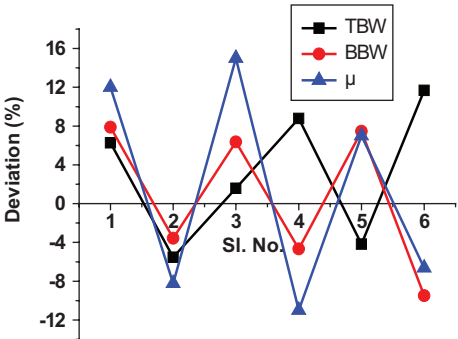


Figure 5.9 Deviation (%) in prediction of TBW, BBW, and μH with respect to their experimental value.

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Chapter 6

Influence of Surface Heat Treatment Performed with Pulsed Nd:YAG Laser on the Gain of Surface Mechanical Properties in VP 100 Tool Steel

G. S. Crespo*, A. C. Gonçalves, V. A. Ventrella

Department of Mechanical Engineering, São Paulo State University (UNESP), School of Engineering of Ilha Solteira, Ilha Solteira, SP, Brazil

Abstract

The literature brings several studies about VP 100 steel, tool steel with a bainitic structure used in the mold, and die manufacturing sector, in the face of machining processes. However, few studies deal with the behavior of this steel against the use, directly or indirectly, of welding processes in its manufacture. Thus, this work analyzed the influence on the structure, and on the surface mechanical properties (hardness and wear rate), of the VP 100 steel when submitted to a surface quenching process with laser. Initially, through the use of a Nd:YAG laser system, in the pulsed condition, six different welding conditions were investigated: SG, AR, N, N1, N2, and N3. The six welding conditions were analyzed using Vickers microhardness tests to determine the best conditions for performing the surface tempering treatment. The three conditions (N1, N2, and N3) that presented the best hardness results, both superficially and in the depth range up to 0.30 mm, advanced in the study. The next step consisted of the execution of the surface quenching process, carried out through the execution of successive autogenous weld beads, in the tempering conditions T1, T2, and T3 (analogous to the welding conditions N1, N2, and N3), on the surface of VP 100 steel proof. The heat-treated specimens

* Corresponding author's email: gillian.engmec@gmail.com.

were analyzed through X-ray diffractograms, and the detection of nitride formation in the post-treatment metal did not occur. Still about the X-ray diffractograms, the data obtained showed a change in the material's structure, from bainitic to martensitic, and a reduction in the size of crystallites in the structures of the three conditions. The wear rate tests showed better results for the harder conditions, with better results being obtained by the T3 condition. Finally, condition T3 was stipulated as the best condition, among those analyzed, for performing surface hardening by laser on VP 100 tool steel, presenting, in addition to an increase in its hardness values, a low wear rate, factors attributed to the formation of a martensitic structure, reduction of crystallite size, and possible nitrogen interstitialization in the metal after heat treatment.

Keywords: Tempering, Toughness, Wear, Nitrogen

Introduction

The plastic injection mold industry is a growing sector due to the widespread use of plastic parts in various sectors of the industry. Such molds are manufactured from cast and rolled steel bars, whose cavities are generated by machining operations, which represent a large part of the manufacturing time and the final price of the mold, so that steel has more influence on the final price of the mold due to its machinability than due to the price of the raw material. Obtaining good parts is directly linked to good molds, since any imperfection will be replicated in the part (Rodrigues et al. 2015). Plastic molds must have mechanical strength combined with hardenability, machinability, polishability, response to texturing, weldability, thermal conductivity, response to surface treatments, and ease of receiving electroerosion process. Nitriding response, weld repair, and corrosion resistance are also important factors (Mesquita 2010). In this scenario, the VP 100 tool steel, developed by Villares Metals, has been gaining great market prominence.

VP 100 is a steel developed with microalloying elements such as Ti and V, which has a lower bainite microstructure in the presence of titanium carbonitride particles. Some of the main known microalloying elements are Nb, Ti, Al, Si, B, and Mo (Medeiros 2011). Titanium forms stable carbonitrides at high temperatures, which contribute to the control of austenitic grain growth prior to thermomechanical processing. Vanadium is the element whose carbonitrides present greater solubility, being normally added to promote the hardening of the ferritic matrix by precipitation (Medeiros 2011).

VP 100 steel can be hardened under different cooling conditions than traditional quenching, presenting homogeneous hardness from the surface to

the core of the block with dimensions of up to $400 \times 1200 \text{ mm}^2$ in section (Medeiros 2011). This steel offers differentiated properties when compared to other steels in the applications of molds, mold holders, and other devices that require high strength, with approximately 32 HRC of hardness, in addition to a good surface quality. Its main competitor in this range of mechanical resistance is the DIN 1.2738 steel of the P20 class, whose similar produced by Villares Metals is the VP20 ISO steel, with which it must be compared in terms of hardness uniformity after heat treatment, machinability, weldability, conductivity, and thermal and residual stresses (Medeiros 2011).

In the traditional steels used in the manufacture of molds, with a hardness of around 35 HRC and a tensile strength of approximately 1030 MPa until breaking, high levels of alloying elements such as Cr, Ni, and Mn are used, which give the steel the ability to hardening by quenching in large sections. This hardenability is essential in deep cavity molds. However, the addition of some of these elements, such as nickel for example, increases the value of the final product. Furthermore, the fact that the cooling does not occur uniformly, from the surface to the core during quenching, leads to the appearance of residual stresses, promoting a non-uniform behavior during machining (Medeiros 2011). In the quest to minimize these effects, VP 100 steel was developed to be hardened with slower cooling speeds than the traditional quenching process, thanks to the effect of microalloying elements such as Ti and V that induce the formation of a lower bainite microstructure, whose hardness is homogeneous for fast and slow cooling (Medeiros 2011).

The lower use of alloying elements in VP 100 steel, such as the low Cr content, contributes to reducing its manufacturing costs, increasing its competitiveness. Its design does not employ high cooling rates, leading to a considerable reduction in residual stresses. In addition to these characteristics, the fact that this steel has a lower carbon content than traditional steels means that, after welding, it presents easier machining adjustment due to the lower hardness generated. The low C content makes the steel less susceptible to hardening caused by rapid cooling in the air after the heat generated by the application of the weld has stopped. To machine it, lower cutting speeds and higher feeds are recommended, and its behavior, in the texturing and polishing processes, is similar to that of traditional steels, such as VP20 ISO (Medeiros 2011).

However, the current industry cannot be limited to the properties of the materials derived from their manufacture so that the search for improvement in the physical–chemical properties of materials is something of great importance in the current research scenario. Among the several ways to change properties, one that has gained great prominence is the use of laser systems, as in the case of the study of surface quenching parameters by laser by diode in the microstructure of tool steels for cold work (Writzl 2011).

Laser surface treatments are now an industrial reality, with several applications, but with the same objective of improving certain surface properties of materials such as mechanics (hardness, fatigue resistance) and tribological (friction, wear), and chemical (corrosion). The perception of these benefits led to the creation of another branch within engineering: surface engineering (Writzl 2011).

The laser surface engineering processes are divided into two large groups: the first being applications that require limited energy/power and do not cause a significant change in phase and/or physical state, and the second that encompasses applications that require a substantial amount of energy to induce phase and/or physical state transformations. The first group includes annealing and etching semiconductors, curing polymers, marking integrated circuit components, and so on. The second group covers cutting, welding, melting, and heat treatments such as surface quenching.

For processes belonging to the second group, laser power/efficiency and interaction time are crucial, as the processes involve single or multiple phase changes within a very short time. In the case of surface quenching, it required lower power density and shorter interaction time between the beam and the material than the fusion and vaporization processes, without reaching fusion (Maillet 1987). The power density used for the laser hardening process is much lower than the power density used for cutting and welding, ranging from 1000 to 2000 W/cm², and occasionally powers much higher as 5000 W/cm² and much lower, as 500 W/cm² (ASM International Handbook Committee 1991). Due to the high energy requirement, for this group of operations, CO₂ and Nd:YAG lasers become the best choice.

The laser beam used in surface engineering processes has the ability to modify the layer of materials to a depth of 0.01–5 mm, depending on the process parameters, although depths of 0.05–1 mm are the most common to be produced (ASM International Handbook Committee 1994). The increase in demand in laser materials processing can be attributed to its unique advantages of high productivity, elimination of a large part of the finishing operation, reduced cost, possibility of process automation, higher quality products, and the generation of a minimum heat-affected zone (HAZ).

Although the VP 100 tool steel presents several works in the literature (Rodrigues et al. 2015; Medeiros 2011; Norcino 2013), regarding its characteristics and behavior in the face of machining processes, this same material suffers a lack of studies regarding its characteristics and behavior in the face of welding processes (or use it). An example of pioneering on this front is the study by Franco et al. (2014) who developed a study where they evaluated VP 100 steel when subjected to TIG welding in the “bead on plate” condition.

Thus, the study developed had the general objective of analyzing the influence resulting from the performance of surface tempering treatment, through the use of a Nd:YAG laser beam, in the pulsed condition, under the physicochemical characteristics of the VP 100 tool steel. Within this general scope, three stages of analysis were carried out. The first was the investigation of the surface hardness, and in a depth range up to 0.30 mm, from the surface of the VP 100 steel, after the superficial heat treatment used. The second encompassed the investigation regarding the structural alteration, and the variation of the average size of the crystallite, through the execution of X-ray diffractograms in the VP 100 steel, after the applied surface heat treatment. The third and last stage focused on the evaluation of the variation in the wear rate of the material by performing a wear test by rotating the ball on VP 100 steel, after the surface heat treatment used.

Methods

The chemical composition of VP 100 tool steel (base metal), as received from the factory, is shown in Table 6.1.

Welding Conditions: SG, AR, and N

Using a United Winners model UW 150A Nd:YAG laser system, in the pulsed condition, autogenous weld beads were performed under three conditions: SG (no shielding gas), AR (argon as shield), and N (nitrogen as shielding gas). Each condition used a specimen made from the base metal, with a square shape with a side of 25.4 mm and a thickness of 6.0 mm. Table 6.2 shows the operational characteristics of the laser system used.

Each specimen (SG, AR, and N) received three weld beads, spaced apart, with no overlap between the beads. Enough time was allowed between the completion of each cord so that there was no thermal influence of the cord performed previously. The parameters used in the welding of each of the three specimens are shown in Table 6.3. Then, the specimens were sectioned, transversely to the welding direction, and hot embedded in Bakelite. The embedded

Table 6.1 Composition of VP 100 tool steel as received from the factory (% by weight)

C	Mn	Si	Mo	Cr	Ni	P	S
0.22	1.90	0.40	0.25	0.58	0.40	0.027	0.002

Table 6.2 Operating characteristics of the Nd:YAG laser system, model UW 150A

Wave length	1064 nm
Maximum laser power	150 W
Peak power	7 kW
Maximum pulse energy	80 J
Pulse width	0.5 ms–30 ms
Frequency	1 Hz–300 Hz
Positioning system	Red laser and CCD camera

Table 6.3 Welding parameters SG, AR, and N

Specimen	SG	AR	N
Shielding gas	-----	Argon	Nitrogen
Welding speed	2 mm/s	2 mm/s	2 mm/s
Frequency	5 Hz	5 Hz	5 Hz
Gas flow	-----	15 l/min	15 l/min
Power	2 kW	2 kW	2 kW
Temporal width	4 ms	4 ms	4 ms
Pulse energy	8 J	8 J	8 J

specimens were then subjected to the conventional metallographic preparation process, with a subsequent attack by Villela reagent to reveal the metallographic structure of the weld beads.

The next step of the study consisted of carrying out Vickers microhardness tests, carried out in a durometer, model DUH-211S, from Shimadzu, using a load of 1 N, with a minimum force of 1.96 mN, speed of 70.0670 mN/s, hold time of 5 s. Hardness tests were carried out on each of the three weld beads of the SG, AR, and N conditions, with the hardness measurement process, with a step between measurements equal to 0.05 mm, divided into three stages: measurements along the entire the surface of each cord; measurements in the horizontal direction, in the central region of each bead; measurements in the vertical downward direction, along the depth of each bead.

Welding Conditions: N1, N2, and N3

Continuing the study, there was a change in the welding parameters of the N condition (chosen due to the hardness results presented), seeking a welding configuration that would allow a better gain of surface mechanical properties

Table 6.4 Welding parameters N1, N2, and N3, compared to welding parameters N

Specimen	N	N1	N2	N3
Shielding gas	Nitrogen	Nitrogen	Nitrogen	Nitrogen
Welding speed	2 mm/s	2 mm/s	2 mm/s	2 mm/s
Frequency	5 Hz	10 Hz	5 Hz	5 Hz
Gas flow	15 l/min	15 l/min	15 l/min	25 l/min
Power	2 kW	2 kW	3 kW	2 kW
Temporal width	4 ms	4 ms	4 ms	4 ms
Pulse energy	8 J	8 J	12 J	8 J

for the VP 100 steel: new welding conditions using nitrogen gas: N1, N2, and N3. The conditions N1, N2, and N3 underwent alterations, respectively, in the following parameters of the welding process: frequency, power, and flow of shielding gas. The other welding parameters were kept constant in relation to the process performed under SG, AR, and N conditions. The welding parameters of conditions N1, N2, and N3 are shown in Table 6.4.

In welding conditions N1, N2, and N3, only one weld bead was performed for each condition. After the welding process, the specimens were sectioned, transversally to the welding direction, and hot embedded with Bakelite, being then subjected to the conventional metallographic preparation process, with subsequent revelation of the metallographic structure of the weld beads through etching, with 5% Nital reagent.

The beads were subjected to Vickers hardness tests, with measurements being carried out in three different cross-sections along the longitudinal direction of each bead, with each section being obtained through the sanding process of the previous section used. The hardness measurement in each section took place with a spacing between measurements equal to 0.05 mm, being performed along the entire surface of the bead and, in a vertical downward direction along the depth of the bead.

After the hardness analysis in the six welding conditions was concluded, a synthesis of the data of all conditions was carried out, both superficially and at a depth of approximately 0.30 mm from the surface of each bead. From this synthesis, the conditions with the best pair of hardness results were chosen to proceed in the following steps developed in the study.

X-ray Diffraction Analysis

The next stage of the study consisted of performing X-ray diffraction analysis, both on the base metal and under welding conditions N1, N2, and N3,

conditions chosen due to the hardness results. For this purpose, four specimens of square dimensions, with 20 mm of side and 2 mm of thickness were made: MB, T1, T2, and T3 (Figure 6.1). The MB specimen was intended for the study of the base metal. This specimen underwent only the conventional metallographic preparation process. The specimens T1, T2, and T3 were subjected to a surface hardening process by laser under the welding conditions used in specimens N1, N2, and N3, respectively.

The tempering process was carried out with the execution of successive autogenous weld beads, parallel to each other, with an overlap rate of 50%, until the surface of each specimen was fully covered (Figure 6.2). Figure 6.3 shows in greater detail one of the three specimens, before and after performing the surface hardening process with a laser beam.

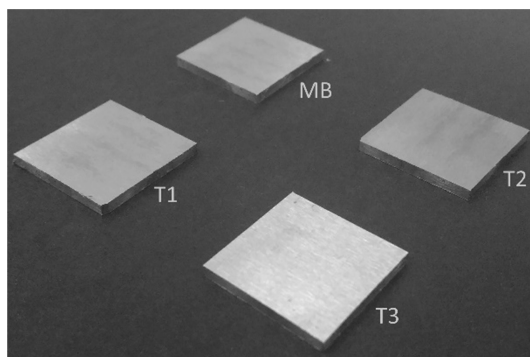


Figure 6.1 Specimens for X-ray diffraction analysis: MB (base metal) and T1, T2, and T3 (surface hardening in welding conditions N1, N2, and N3).

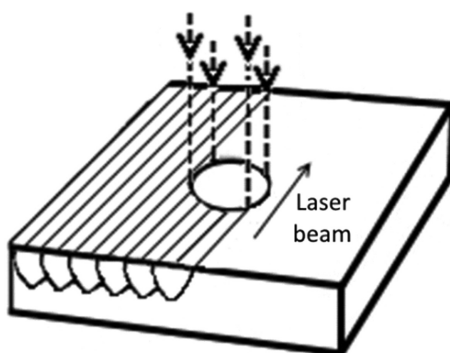


Figure 6.2 Representative diagram of the surface hardening process by laser.

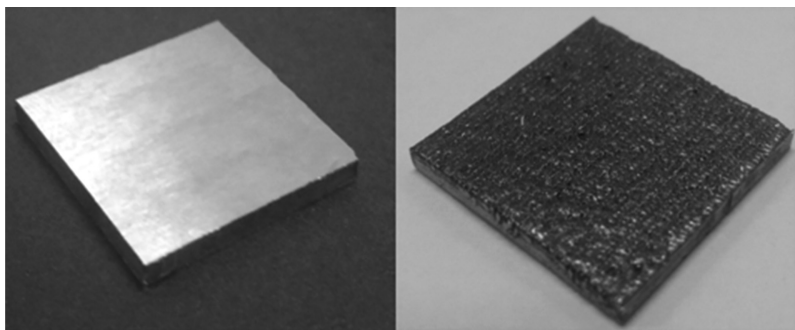


Figure 6.3 Specimen submitted to the surface hardening process by laser: (left) before the process and (right) after the process.

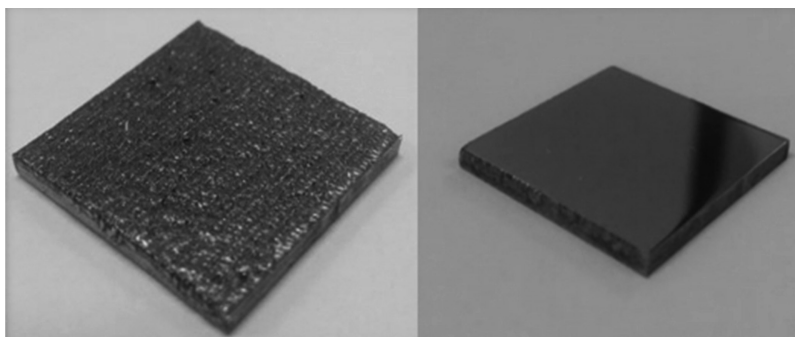


Figure 6.4 Specimen analyzed in the X-ray test: (left) directly from the surface hardening process by laser and (right) after the conventional metallographic preparation process.

The analyses were carried out in an X-ray diffractometer, model XRD-6000, from Shimadzu, located in a room maintained at 23°C, 24 h a day. The analyses used an aluminum sample holder and a wavelength $\lambda = 1.54056 \text{ \AA}$, with a sweep of 1°/min in the angular range from 5° to 100°. Each specimen was subjected to X-ray analysis under two conditions (Figure 6.4): directly originating from the surface laser hardening process, without external alteration of the surface structure obtained as a result of the hardening process, and prepared by conventional metallography process, thereby obtaining a smooth, uneven surface laser hardened surface. The test under these two conditions had the objective of analyzing the variation of the material's structure, after the surface quenching process, along the depth from the surface. In the case of the MB specimen, the test was performed only once, and the results obtained for it were used as a basis for comparison and control of the results obtained in the specimens that underwent the surface hardening process by laser.

The results of the X-ray diffractograms obtained were used for the determination of the material composition after the completion of the surface quenching process by laser (investigation of phase alteration in the base material) and investigation of the possible formation of nitrides, resulting from the interaction of the nitrogen gas used with the weld pool, during the surface quenching process by laser.

Calculation of Crystallite Size

The next step of the study consisted of calculating the average size of the crystallite, using the Scherrer equation (Equation 6.1), for welding conditions N1, N2, and N3.

$$t = \frac{k \cdot \lambda}{B \cdot \cos(\theta_B)} \quad (6.1)$$

where

- t : average crystallite diameter (Å)
- k : constant that varies according to the geometry of the particles
- λ : wavelength of electromagnetic radiation (Å)
- θ_B : diffraction angle (degrees)
- B : width at half the height of the diffraction peak (radians)

The calculations were developed using data from the X-ray diffractograms developed in the specimens T1, T2, and T3, in the two conditions of realization of the diffractograms. The purpose of this step was to investigate the change in the size of the crystallites in each condition, verifying the occurrence of refinement in the structure of the material, after the surface hardening process by laser.

Calculation of the Wear Rate

In the final part of the study, microwear tests were carried out using a fixed rotating ball, in the base metal and in the tempering conditions T1, T2, and T3, to calculate the wear rate in each condition. The tests were carried out in a microwear test machine using a fixed rotating ball, using an AISI 52100 steel ball (quenched and tempered), with a diameter of 25.4 mm (1 in). For each tempering condition, two conditions (Table 6.5) of the microwear test were performed.

Table 6.5 Parameters used in wear tests, micro-abrasive by fixed rotating ball, in tempering conditions T1, T2, and T3

Condition	1	2
Time [s]	300	900
Normal force [N]	4.9	4.9
Sphere rotation [rpm]	300	300
Slip distance [m]	119.7	359.07
Number of repetitions	3	3

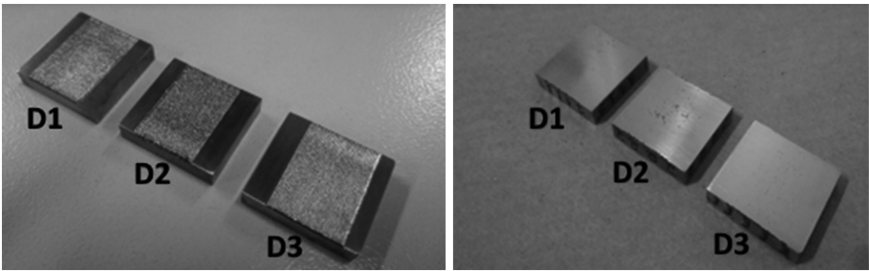


Figure 6.5 Specimens D1, D2, and D3: (left) after laser surface hardening process and (right) after sanding process.

The tests were carried out on specimens with square dimensions of 25.4 mm side and 12.5 mm thickness. The MB specimen, intended for the study of the base metal, underwent only a grinding process with subsequent sanding with #120 grit sandpaper. The specimens D1, D2, and D3 were subjected to a surface hardening process by laser in the conditions of surface hardening by laser T1, T2, and T3, respectively, with a subsequent sanding process with sandpaper of #120 granulometry. The tempering process was carried out with the execution of successive autogenous weld beads, parallel to each other, with an overlap rate of 50%. Figure 6.5 shows samples D1, D2, and D3, after performing the surface hardening process and after the sanding process.

After performing the microwear tests, the samples were photographed under a common optical microscope to obtain images of the wear caps resulting from the tests. For each cap obtained, a spherical cap was determined, and the diameter (d) of this spherical cap was obtained through computerized measurement using the ImageJ image processing software. Each spherical cap had its diameter measured twice: vertically and horizontally. From the diameters obtained, the volumes (V) of the wear craters (removed material)

were calculated using Equation 6.2, in accordance with the ISO 26424:2008 standard.

$$V = \frac{\pi \cdot d^4}{64 \cdot R} \quad (6.2)$$

For $d \ll R$, where d is the diameter of the spherical cap obtained in the test and R is the radius of the sphere used in the test.

In the next step, from the values of V obtained, the wear rates (Q) were calculated using Equation 6.3.

$$Q = \frac{V}{S} \quad (6.3)$$

where S represents the test slip distance (m).

Results and Discussion

Weld Metal Hardness: SG, AR, and N

Figure 6.6 shows the three weld beads performed under SG, AR, and N welding conditions. The indentations of the hardness tests performed can also be seen in the images. Table 6.6 provides the global mean surface hardness values for the weld metals in SG, AR, and N conditions, together with the base metal global mean hardness value. Table 6.7 shows the surface hardness gains of each condition, calculated in relation to the average global hardness of the base metal.

From Table 6.6, although the difference presented between the AR and N conditions is not very pronounced, this difference exists, with the N condition presenting less deviation of results than the AR condition. The presence of inert or active shielding gas also generated, when compared to the SG condition, higher surface hardness, probably due to the high cooling rate imposed on the weld pool by the shielding gas flow.

In the tempering process, the higher the cooling rate used in the heated region, the greater the chances of forming structures that have greater hardness, commonly martensite. Thus, it can be conjectured that the bainitic structure of the base metal probably underwent a heating process until austenitization, and that the presence of the shielding gas was a major influencer in the generation of a higher concentration of martensitic structure in the AR and N conditions, in relation to the SG condition.

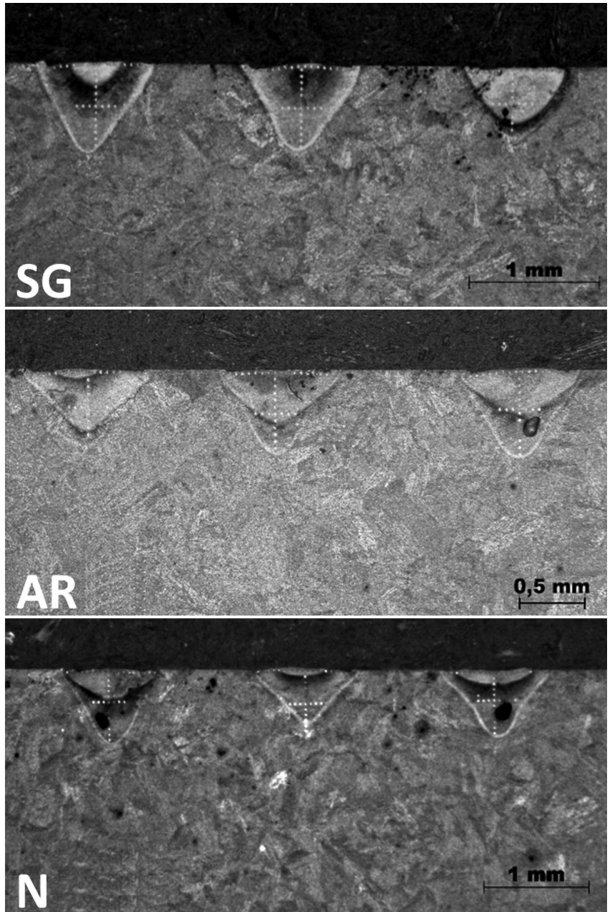


Figure 6.6 Weld beads performed under SG, AR, and N conditions. Attack: Villela reagent.

Table 6.6 Overall mean hardness values for the base metal and for the surfaces of weld metals under SG, AR, and N conditions

Condition	MB	SG	AR	N
Global average hardness	359.2	482.6	590.2	630.7
±	±	±	±	±
Standard deviation	22.4	89.4	103.3	94.6

Values in HV.

Table 6.7 Surface hardness gain values, in relation to the average global hardness of the base metal, of the SG, AR, and N welding conditions

Condition	SG	AR	N
Gain	34.4%	64.3%	75.6%

Table 6.8 Global average hardness values for the base metal and for the central regions of the weld metals under SG, AR, and N conditions

Condition	MB	SG	AR	N
Global average hardness	359.2	575.6	614.4	586.8
±	±	±	±	±
Standard deviation	22.4	37.2	59.3	84.3

Values in HV.

The lower surface hardness of the SG condition can also be evaluated as a result of possible surface decarburization. Decarburization would occur due to the interaction of the carbon present in the steel with the free oxygen of the environment, which would lead to the loss of carbon from the steel from the surface, generating a layer with reduced carbon content, consequently presenting lower hardness, since the hardness of martensite directly depends on the carbon content. Thus, under AR and N conditions, in addition to the cooling role, the gas flow used would also have played the role of protecting the surface against oxidation resulting from the interaction with the medium, thus avoiding this possible decarburization of the surface, reflecting in hardness, greater than those present in the SG condition.

It was also inferred that, since the conditions that used shielding gas were performed using the same parameters, nitrogen gas (active gas) played a role beyond the simple imposition of a high cooling rate in the weld pool when compared to argon gas (inert gas). Thus, it is conjectured that the nitrogen gas may have acted in two possible ways: either in the formation of some nitride in the structure of the weld metal or, with the nitrogen being incorporated into the matrix of the material, in an interstitial way, in the form of a solution. supersaturated solid.

Table 6.8 shows the global average central hardness values for each of the conditions, together with the global average hardness value of the base metal. Table 6.9 shows the global average central hardness gains for each condition, calculated in relation to the global average hardness of the base metal.

From Table 6.8, it is evident that the difference in hardness obtained, when it comes to the horizontal central region of the weld bead, between the three

Table 6.9 Central hardness gain values, in relation to the average global hardness of the base metal, of the welding conditions SG, AR, and N

Condition	SG	AR	N
Gain	60.2%	71.0%	63.4%

Table 6.10 Global mean hardness values for the base metal and for the central descending regions of the weld metals under SG, AR, and N conditions

Condition	MB	SG	AR	N
Global average hardness	359.2	537.5	582.8	559.4
±	±	±	±	±
Standard deviation	22.4	105.0	108.7	132.8

Values in HV.

Table 6.11 Central descending hardness gain values, in relation to the average global hardness of the base metal, of the SG, AR, and N welding conditions

Condition	SG	AR	N
Gain	49.6%	62.2%	55.7%

conditions, was not as discrepant as that found on the surface of the beads. This can be attributed to the fact that, from the moment the surface boundary of the molten pool solidified, the cooling mechanism for all conditions assumed a very similar character so that all conditions ended up reaching a plateau of very close final hardness. Comparing the conditions with the presence of gas, it is noted that this time there was a higher average hardness presented by the argon gas. One inference is that the difference in hardness gain of nitrogen gas in relation to argon gas is something that is limited to the surface region of the weld metal, and that it is not maintained when dealing with other regions of the weld bead.

Table 6.10 brings the global mean central descending hardness values for each of the conditions, together with the global mean hardness value of the base metal. Table 6.11 provides the global average descending central hardness gains for each condition, calculated in relation to the global average hardness of the base metal.

Close to the surface, the hardness presented by the conditions with the presence of gas was higher than those presented by the condition without the

presence of gas, probably due to the high cooling rates imposed by the flow of gas present in the process. However, from a certain depth, this differentiation suffered a decay. This fact, as well as what happened in the horizontal central position of the beads, is probably linked to the fact that, once the surface boundary of the weld beads is solidified, the cooling mechanism becomes equal for all conditions, which, consequently, levels hardness value thresholds for conditions with and without the use of shielding gas.

Weld Metal Hardness: N1, N2, and N3

Figure 6.7 shows the three measurement sections in N1, N2, and N3 welding conditions. The indentations of the hardness tests performed can also be seen in the images.

Table 6.12 shows the global average surface hardness values for the weld metals of each condition presented, together with the global average surface hardness value of the N condition, for comparison between all conditions

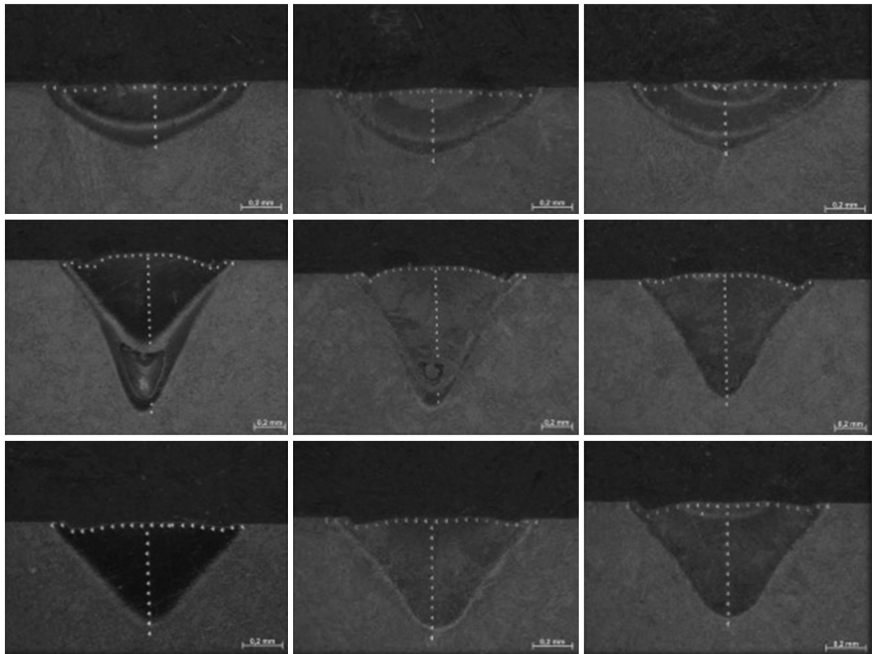


Figure 6.7 Measurement sections of weld beads N1, N2, and N3. Lines (from top to bottom): conditions N1, N2, and N3. Columns (from left to right): sections 1, 2, and 3. Attack: Nital 5%. Scale: 0.2 mm.

Table 6.12 Global average surface hardness values for conditions N, N1, N2, and N3

Condition	N	N1	N2	N3
Global average hardness	630.7	633.2	678.1	668.7
±	±	±	±	±
Standard deviation	94.6	67.2	89.0	86.1

Values in HV.

Table 6.13 Percentage gains in surface hardness of N1, N2, and N3 conditions, when taking as a basis the global average hardness of the base metal and surface of N condition

Condition	Comparison with MB	Comparison with condition N
N1	76.3%	0.4%
N2	88.8%	7.5%
N3	86.2%	6.0%

using nitrogen gas. Table 6.13 shows the percentage gains in surface hardness of the N1, N2, and N3 conditions, when based on the global average hardness of the base metal and surface hardness of the N condition.

Although all three conditions presented the high surface hardness values, the N2 and N3 conditions presented higher values than the N1 condition. Based on the surface hardness of the N condition as a basis for comparison, the variation of parameters applied in the N1, N2, and N3 conditions was not all of the effective action for a significant increase in the surface hardness of the weld metals. The frequency variation applied in the N1 condition generated a surface hardness equivalently identical to the hardness obtained on the surface of the weld metal performed in the N condition. The N2 and N3 conditions showed an increase in relation to the N condition.

In the case of the N2 condition, it can be inferred that the higher hardness presented is due to the higher pulse energy used, which consequently generated a higher temperature, allowing a more efficient austenitization of a greater volume of material, favoring the formation of martensite. In this scenario, it can be argued that a higher temperature provided a better rate of “martensitization” of the base metal structure in addition to, consequently, a longer time for nitrogen gas interaction with the weld metal pool, which would generate more time for either nitrogen interstitialization in the weld metal structure, or for the formation of possible nitrides in the weld metal structure.

In the case of the N3 condition, it is possible to infer the highest hardness presented with an analysis similar to that used in the N2 condition. The higher flow of shielding gas used consequently generated higher cooling rates, which in turn positively favored the “martensitization” of the structure in addition to, consequently, a higher flow of gas providing the weld metal with a greater amount of nitrogen for interaction., either for nitrogen interstitialization in the weld metal structure or for the formation of possible nitrides in the weld metal structure.

Regarding the hardness in the central downward direction, since the weld bead performed in the N1 condition had a lower average depth than the beads performed in the other conditions that used nitrogen gas, for an effective comparison, only the initial measurements of the N, N2, and N3 conditions, up to the maximum value of measurements developed in the N1 condition and that have been developed in the three adopted measurement sections. Thus, all comparisons were performed up to depth measurement number 6.

Table 6.14 shows the global mean central descending hardness values for the first six measurements of conditions N, N1, N2, and N3. Table 6.15 shows the percentage gains in global average descending hardness of the first six depth measurements of conditions N1, N2, and N3, when taken as a basis for

Table 6.14 Global average central descending hardness values for the first six measurements of conditions N, N1, N2, and N3

Condition	N	N1	N2	N3
Global average hardness	527.6	515.5	581.2	602.3
±	±	±	±	±
Standard deviation	142.2	79.5	62.2	63.1

Values in HV.

Table 6.15 Percentage gain values of the global average central descending hardness for the first six measurements of the weld metals in the N1, N2, and N3 conditions, when based on the values of the global average base metal hardness and the global average central descending hardness of the first six N condition measures

Condition	Comparison with MB	Comparison with condition N
N1	43.5%	−2.3%
N2	61.8%	10.2%
N3	67.7%	14.1%

the global average hardness of the base metal and, central descending of the first six depth measurements of the condition N.

Both the N2 and N3 conditions showed higher hardness values than those presented by the N1 condition, establishing that the frequency change performed in the N1 condition did not result, again, in any significant gain in the hardness property when compared to its base condition.

Summary of Hardness Results

Table 6.16 shows the global mean surface hardness gain values for SG, AR, N, N1, N2, and N3 conditions when based on the global mean hardness of the base metal.

Comparing all the surface hardness data, it appears that, from the change in the parameters of the N condition, which had already presented the best performance among the first conditions analyzed, an even greater gain in surface hardness was possible from the base metal, highlighting the conditions N2 and N3 competing as the best conditions among those analyzed. The analysis of results in the center of the cords in the SG, AR, and N conditions showed an uninteresting scenario where, from a certain depth, the conditions start to compete with each other, with not such a marked difference between conditions with and without, the use of shielding gas, whether active or inert. Since the study was interested in the gain of surface properties and in a depth range of up to 0.30 mm, it was decided to use the analysis of surface hardness and, central descending, to choose the best conditions that would continue to be analyzed. in the study. Table 6.17 brings the values of the global average descending central hardness gain, for the first six depth measurements, approximately 0.30 mm, for the conditions SG, AR, N, N1, N2, and N3, when based on the hardness global average of the base metal.

Analyzing Table 6.17, it is inferred that at the analyzed depth, the conditions that showed greater central descending hardness were the AR and N3 conditions, followed very closely by the N2 condition. Condition N1 again did not stand out in relation to the value obtained by condition N, even presenting

Table 6.16 Values of percentage gains of the global average surface hardness for the conditions SG, AR, N, N1, N2, and N3, when based on the global average hardness value of the base metal

Condition	SG	AR	N	N1	N2	N3
Gain	34.4%	64.3%	75.6%	76.3%	88.8%	86.2%

Table 6.17 Values of percentage gains of the global average central descending hardness, up to a depth of 0.30 mm, for the conditions SG, AR, N, N1, N2, and N3, when based on the global average hardness value of the base metal

Condition	SG	AR	N	N1	N2	N3
Gain	39.6%	67.3%	46.9%	43.5%	61.8%	67.7%

a value lower than its base condition. Due to this fact, it was decided to continue the study with those conditions that reconciled greater surface hardness, allied to good values of central descending hardness up to a depth of 0.30 mm. Within the adopted criterion, the conditions N2 and N3 were chosen, which presented the highest surface hardness, and the N3 condition also presented the highest central descending hardness up to 0.30 mm. Although the AR condition presented a higher central descending hardness up to 0.30 mm than the N2 condition, the difference presented was considered small, and since the surface hardness gain of the N2 condition was very accentuated compared to that presented by the N2 condition, condition AR, we opted for condition N2. Even having presented a considerably lower value in relation to the AR condition, the N1 condition was also taken for further study, due to its higher surface hardness value, which, compared to the AR condition, was considered interesting to the point of being also included in the following analyses developed in the study.

X-ray Diffraction Analysis

Figure 6.8 shows the X-ray diffractograms obtained for the base metal, after the conventional metallographic preparation process and, for the surface temper conditions T1, T2, and T3, on the surfaces directly arising from the process.

Performing the image analysis, the peaks presented by the base metal are attributed to the ferritic structure, a result already expected since the base metal consists of a bainitic structure, which in turn is partially composed of ferrite. The comparison of the peaks visualized in the surface quenching in the conditions T1, T2, and T3 formed, on the surface from the laser quenching process, a martensitic structure. This corroborates the analyses carried out in the hardness stage, where the connection was made that the hardness increases found were related to a possible scenario where the bainitic structure of the base metal would have suffered a temperature rise to the austenitization temperature level and, later, as a result of the high cooling rates, this structure would have been transformed into a martensitic structure, consequently having

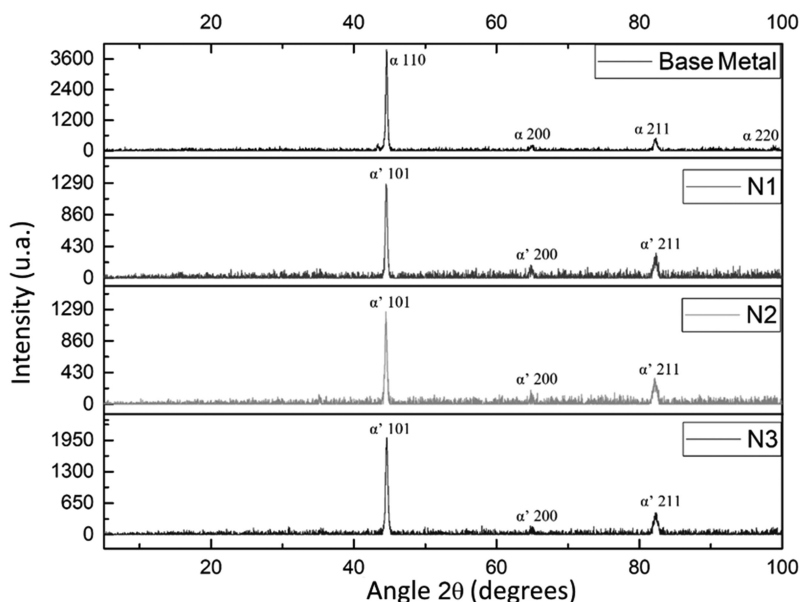


Figure 6.8 X-ray diffractograms for the base metal and for the surface quenching conditions T1, T2, and T3, on the surfaces directly originating from the surface quenching process.

greater hardness than that presented by the base metal. As for the interaction of nitrogen gas with the weld pool, none of the diffractograms identified peaks that corresponded to the formation of some nitride in the weld metal structure. Figure 6.9 shows the X-ray diffractograms obtained for surface quenching (after a conventional metallographic preparation process) carried out under conditions T1, T2, and T3, in addition to the base metal diffractogram.

The comparison of the peaks visualized in Figure 6.9 with the peaks visualized in Figure 6.8 shows that the surface quenching in the conditions T1, T2, and T3, formed, once again, on the surface below the surface directly arising from the process, a martensitic structure, corroborating more once, the analyses carried out in the hardness stage. As for the interaction of nitrogen gas with the weld pool, once again, in none of the diffractograms performed were peaks identified that corresponded to the formation of some nitride in the structure of the weld metal.

Calculation of Crystallite Size

Figure 6.10 shows the X-ray diffractograms for the base metal and, for the surfaces directly originating from the surface quenching process of the surface quenching conditions T1, T2, and T3, superimposed on a single axis, with

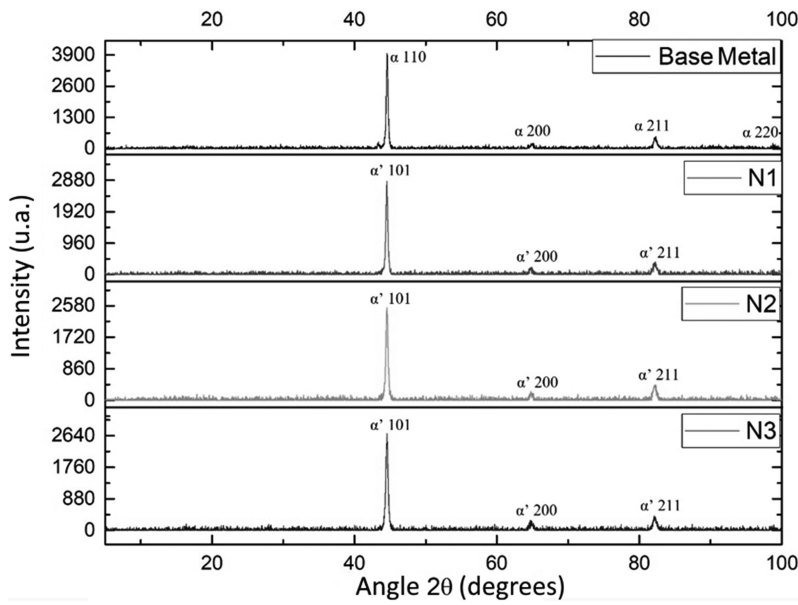


Figure 6.9 X-ray diffractograms for the base metal and, for the surface hardening conditions T1, T2, and T3, after the conventional metallographic preparation process.

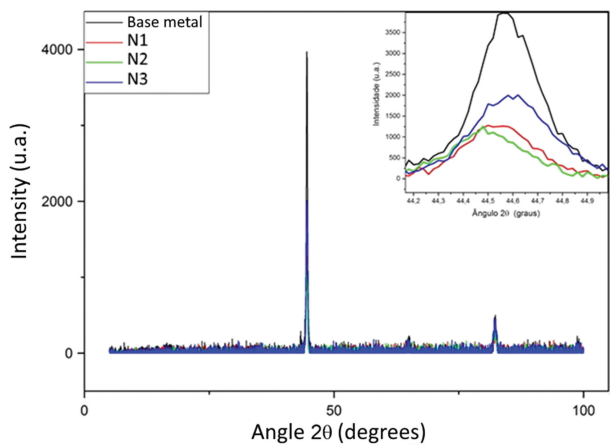


Figure 6.10 X-ray diffractograms of the base metal and surface quenching conditions T1, T2, and T3 (surfaces directly derived from the quenching process), superimposed on a single axis. (Detail) Magnification of the first peak region (~44.55°), showing the region with the greatest variation between the analyzed conditions.

Table 6.18 Parameters used in the calculation, together with the calculated value, for the crystallite size, through the Scherrer equation, and the reduction rate in relation to the value obtained for the base metal, for the surface hardening conditions T1, T2, and T3 (surfaces directly from the surface tempering process)

Condição	k (cte)	λ (Å)	B (rad)	θ (°)	t (Å)	Reduction (%)
Base Metal	0.9	1.5413	0.0052	44.44	371	0
T1	0.9	1.5413	0.0061	44.51	318	14.2
T2	0.9	1.5413	0.0056	44.55	349	6.1
T3	0.9	1.5413	0.0065	44.61	302	18.7

detail focusing on the variation that occurred in the first peak ($\sim 44.55^\circ$). This focus is due to the fact that the application of the calculation of the crystallite size variation, through the Scherrer formula, is performed at the peak where there was the greatest variation between the diffractograms to be analyzed.

Table 6.18 presents the parameters used in the execution of the Scherrer formula for each condition, together with the results of the average crystallite size (Equation 6.1) and the percentage of reduction suffered in each condition, compared to the value calculated for the crystallite of the base metal.

From Table 6.18, we have that the greatest reduction in crystallite size occurred in the T3 condition, followed by the T1 and T2 conditions, respectively. When talking about crystallite size, two factors must be considered: high temperature and residence time at high temperatures. Higher temperatures, combined with a longer residence time, tend to generate larger crystallite sizes. The size of the crystallites is in turn closely linked to the hardness of the material. A more refined structure (composed of smaller crystallites) has greater hardness than a coarser structure (composed of larger crystallites), due to the fact that such a configuration leads to a better distribution of the stresses suffered by the material.

The T1 condition presented one of the highest crystallite size reduction rates, not presenting, however, one of the highest hardness found. The increase in frequency means that the energy used in the process will be delivered in the same time interval; however, this delivery will be performed by a greater number of pulses which, for the same amount of energy, consequently leads to smaller pulses. energy. It can be analyzed with this that this increase in the frequency of the pulses did not provide enough time, at high temperatures, for high growth of crystallites to occur. However, this fact also did not allow an austenitization as high as the other conditions, so that even with a refining in the structure, this refining did not translate into an equivalent increase in hardness levels, when compared to the other conditions.

On the other hand, the T2 condition was, at the same time, the condition that presented the lowest crystallite size reduction rate, together with one of the highest surface hardness found. With the increase in energy, for the same flow of shielding gas, the weld pool will take longer to complete its cooling process. Comparing conditions T1 and T3, which used the same flow of shielding gas, it can be said that condition T2, by presenting higher temperatures in the molten pool, developed a better “martensitization” of the structure, which reflected in high levels of hardness. However, the time for crystallite growth was longer than that provided by the T1 condition, reflecting a coarser structure. It can also be inferred that, since the T2 condition presents one of the highest hardness, even presenting a coarse structure in relation to the T1 and T3 conditions, there is a possibility of nitrogen gas acting on the results of this condition. This performance can be attributed to the fact that a higher pulse energy allowed a longer interaction time between the shielding gas and the weld pool, allowing the occurrence of nitrogen interstitialization in the weld metal, which would justify the high hardness values, even with a low refinement of the structure.

Finally, the T3 condition was presented as the condition that showed the best pair of properties, combining one of the highest hardness found on the surface, together with the highest crystallite size reduction rate generated in the surface layer of the weld pool. The higher flow of shielding gas translates into a higher rate of cooling, reflected in the lower growth of crystallites, which translates into the more refined structure obtained. However, this higher rate of cooling would also lead to lower temperatures reached by the molten pool. Since the energy used by the T3 condition is the same present in the T1 condition, a greater refinement of the T3 structure was expected when comparing the two conditions, given the higher cooling rate of the T3 condition. Thus, it can be inferred that the “martensitization” of the structure seen in the T3 condition would not have been so efficient, so that a level of hardness similar, or even lower, to that presented by the T1 condition should have been obtained, a fact not occurred. Therefore, there is one more condition where there may be the possibility of a decisive role for nitrogen gas. A greater flow of shielding gas probably allowed a greater interaction between the shielding gas and the molten pool, so that, even if the imposed cooling rate had been higher, the greater amount of gas used could have allowed the occurrence of nitrogen interstitialization in the weld metal, which would justify the better hardness values of the T3 condition, compared to the T1 condition, even with the highest cooling rate and other identical parameters.

Figure 6.11 shows the X-ray diffractograms for the base metal and the surface hardening conditions T1, T2, and T3 (surfaces after the conventional metallographic preparation process), superimposed on a single axis, with detail

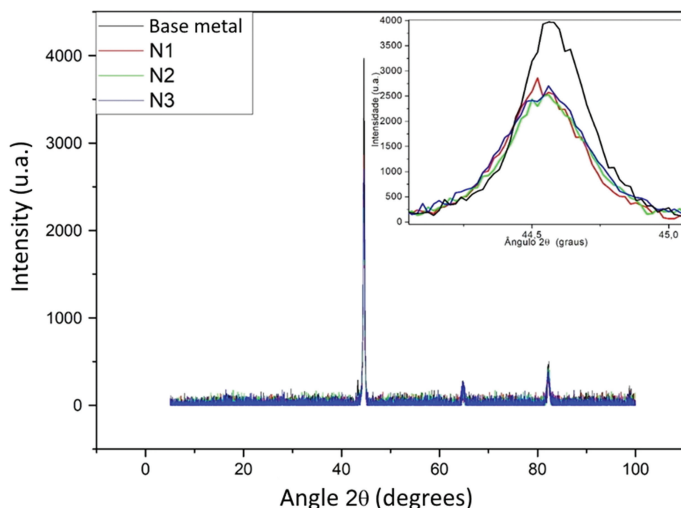


Figure 6.11 X-ray diffractograms of the base metal and surface hardening conditions T1, T2, and T3 (surfaces after conventional metallographic preparation process), superimposed on a single axis. (Detail) Magnification of the region of the first peak ($\sim 44.56^\circ$), showing the region with the greatest variation between the conditions analyzed.

focusing on the variation that occurred in the first peak ($\sim 44.56^\circ$), between the conditions analyzed, following the same analysis passes performed in Figure 6.11.

Table 6.19 brings the parameters used in the execution of the Scherrer formula (Equation 6.1) for each condition, together with the results of the average crystallite size (Equation 6.1) and the percentage of reduction suffered in each condition, compared to the value calculated for the base metal crystallite.

From the values presented in Table 6.19, it can be seen that condition T1, contrary to what occurred in the analysis performed on the surface directly from the quenching process, this time presented the lowest rate of crystallite size reduction. This fact is perfectly linked to the discussed data about hardness in the central region of the beads, allowing the analysis that this was due to the lower crystallite size reduction rate obtained in this region, compared to the value obtained in the calculation performed on the surface directly originating from. of the surface hardening process, which directly resulted in the reduction of the hardness value obtained. As for the T2 condition, contrary to what was observed in the analysis of the results obtained on the surface directly from the surface quenching process, this time there was a full agreement between the calculated crystallite reduction rate and the hardness values

Table 6.19 Parameters used in the calculation, together with the calculated value, for the crystallite size, through the Scherrer equation, and the reduction rate in relation to the value obtained for the base metal, for the surface hardening conditions T1, T2, and T3 (surfaces after conventional metallographic preparation process)

Condição	k (cte)	λ (Å)	B (rad)	θ (°)	t (Å)	Reduction (%)
Base Metal	0.9	1.5413	0.0052	44.44	371	0
T1	0.9	1.5413	0.0054	44.58	360	3.0
T2	0.9	1.5413	0.0059	44.51	328	11.7
T3	0.9	1.5413	0.0063	44.58	310	16.5

obtained. In this case, the hardness values are only correlated with the crystallite reduction rate and cooling phenomena that provide the “martensitization” of the structure, since once the surface boundary has solidified, contact with the shielding gas ceases, with the interruption of the interaction, which, consequently, does not provide an environment that allows the interstitialization of nitrogen in the structure of the weld metal.

Finally, condition T3 was the condition that presented the best results on the surface, after the conventional metallographic preparation process, combining the highest hardness found in the central descending analysis over the first six measurements in each condition, together with the highest reduction rate of crystallite size in the same region. As in the T2 condition, the hardness values are only correlated with the crystallite reduction rate and cooling phenomena that provide the “martensitization” of the structure, since once the surface boundary has solidified, contact with the carbon dioxide gas ceases. protection, with the interruption of the interaction, which, consequently, does not provide an environment that allows the interstitialization of nitrogen in the structure of the weld metal.

Wear Rate

Tables 6.20–6.22 provide, respectively, the values of d (wear crater diameter), V (wear crater volume), and Q (wear rate), for the base metal and for the surface hardening conditions T1, T2, and T3, for each measurement, in each repetition, for each wear test condition performed. Table 6.23 shows the average values for d , V , and Q , for the base metal and for the surface hardening conditions T1, T2, and T3, for each wear test condition performed.

Table 6.20 Values of d (mm), for each measurement, of each repetition, for each wear test condition performed, on the base metal and in the surface hardening conditions T1, T2, and T3

	300 s						900 s					
	1-1	1-2	2-1	2-2	3-1	3-2	1-1	1-2	2-1	2-2	3-1	3-2
Base Metal	1.590	1.691	1.805	1.891	1.039	1.032	2.192	2.242	1.354	1.454	1.439	1.522
T1	1.190	1.186	1.908	1.941	1.057	1.087	2.107	2.180	1.444	1.488	1.573	1.606
T2	1.227	1.234	1.484	1.532	1.680	1.687	1.657	1.694	1.669	1.713	1.783	1.812
T3	1.190	1.208	0.913	0.961	0.702	0.722	1.316	1.358	1.392	1.409	1.120	1.132

Table 6.21 V values ($\text{mm}^3 \times 10^{-4}$), for each measurement, of each repetition, for each wear test condition performed, on the base metal and in the T1, T2, and T3 surface hardening conditions

	300 s						900 s					
	1-1	1-2	2-1	2-2	3-1	3-2	1-1	1-2	2-1	2-2	3-1	3-2
Base Metal	247	316	410	494	45	44	892	977	130	173	166	207
T1	78	76	512	549	48	54	762	873	168	189	237	257
T2	88	90	187	213	308	313	291	318	300	333	391	417
T3	78	82	27	33	9	11	116	131	145	152	61	63

Table 6.22 Q values ($\text{mm}^3/\text{m} \times 10^{-7}$), for each measurement, of each repetition, for each wear test condition performed, on the base metal and in the surface hardening conditions T1, T2, and T3

	300 s						900 s					
	1-1	1-2	2-1	2-2	3-1	3-2	1-1	1-2	2-1	2-2	3-1	3-2
Base Metal	2064	2640	3427	4129	376	366	2485	2719	362	481	462	578
T1	648	639	4279	4583	403	451	2121	2431	468	528	659	716
T2	732	749	1566	1779	2572	2615	811	886	835	927	1088	1160
T3	648	688	224	275	78	88	323	366	404	424	169	177

Table 6.23 Mean values (mean $\pm$ standard deviation) of d , V , and Q , for each wear test condition performed on the base metal and in the surface hardening conditions T1, T2, and T3

300 s			
	d [mm]	V [$\text{mm}^3 \times 10^{-4}$]	Q [$\text{mm}^3/\text{m} \times 10^{-7}$]
Base metal	1.508 ± 0.380	259 ± 0.019	2167 ± 0.000156
T1	1.395 ± 0.414	220 ± 0.024	1834 ± 0.000202
T2	1.474 ± 0.205	200 ± 0.010	1669 ± 0.000083
T3	0.949 ± 0.219	40 ± 0.003	334 ± 0.000027
900 s			
	d [mm]	V [$\text{mm}^3 \times 10^{-4}$]	Q [$\text{mm}^3/\text{m} \times 10^{-7}$]
Base metal	1.701 ± 0.404	424 ± 0.040	1181 ± 0.000111
T1	1.733 ± 0.324	414 ± 0.032	1154 ± 0.000088
T2	1.721 ± 0.063	342 ± 0.005	951 ± 0.000014
T3	1.288 ± 0.129	112 ± 0.004	311 ± 0.000011

Correlating the data obtained for the wear rates with the data obtained in the hardness analysis stage, one has that the best hardness results (T2 and T3 conditions, both superficially and within the analyzed depth range), also showed the lowest, and therefore the best, wear rate values. Among these two conditions, however, condition T3 stands out for the fact that, in addition to presenting the lowest wear rate, the value presented is very far from the values presented by the other conditions. This difference is probably linked to the fact that the structure of condition T3 is the most refined, among the conditions analyzed, both superficially and up to a depth of 0.30 mm, since more refined structures have higher hardness, consequently presenting lower rates of wear.

In the crystallite size analysis, the structure of the T2 condition showed an increase in the reduction rate according to the increase in depth in relation to the surface boundary of the welding process, including a lower value of the reduction rate on the surface than presented by the T1 condition. This can be correlated with the values of wear rate visualized, where superficially the conditions T2 and T1 present similar values and, as the test continues and the depth increases, the wear rate value presented by the condition T2 becomes distance and suffers an abrupt drop when compared to itself at the beginning of the test, a fact less evident in the T1 condition.

Still in the crystallite size analysis, the T1 condition had a high rate of reduction in the size of the structure on the surface and, consequently, a drop in value within the analyzed depth, reaching a value of only 3%. This can also be correlated with the values of wear rate visualized, where the values presented by the T1 condition, and by the base metal, start to approach during the test time.

Finally, Table 6.24 shows the percentage values of reduction of Q values, presented by the surface hardening conditions T1, T2, and T3, when compared with the values presented by the base metal, in each condition of wear test performed.

Table 6.24 Comparative values, of percentage reduction, between the Q values in the T1, T2, and T3 surface hardening conditions, when the Q value for the base metal is taken as a basis

Condition	Reduction (%) – 300 s	Reduction (%) – 900 s
Base metal	0	0
T1	15.37	2.29
T2	22.98	19.48
T3	84.59	73.67

Conclusion

Analyzing all the welding conditions studied, the best surface hardness gain values were presented by the conditions that used nitrogen gas as a shielding gas, and in some cases, the surface hardness gain, in relation to the base metal, reached almost 89% (88.8% and 86.2% for N2 and N3 conditions, respectively). The hardness behavior in the downward vertical direction showed that, from a certain depth, the values presented by the conditions start to have little distance between them. This fact can be seen as a consequence that, from the solidification of the surface boundary of the weld pool, the cooling of the weld metal starts to depend solely on the volume of base metal around it and, on the superiorly solidified weld metal, equalizing the cooling mechanism in all situations, with this an equivalence of values between the conditions. However, considering a hardness variation close to the surface (about 0.30 mm), the conditions that used nitrogen gas as a shielding gas are again highlighted, presenting gains of almost 68% in relation to the values of the base metal.

The analyses by performing X-ray diffractograms, under the surface hardening conditions T1, T2, and T3 (analogous to the conditions N1, N2, and N3), did not show the presence of any nitride formation, both when analyzing the surface directly from the laser surface treatment processes and when analyzing after the conventional metallographic surface preparation process. However, the diffractograms indicated the transformation of the bainitic structure of the base metal into a martensitic structure, which would justify the higher hardness obtained in the analyses.

Regarding the calculations of the crystallite size reduction rate in the T1, T2, and T3 surface quenching conditions, all conditions showed crystallite size reduction rates (structure refinement), with the highest reduction rates presented by the T3 condition. This refinement in the structure is reflected in the hardness values obtained, so that the best rates of structure refinement also presented the best hardness values.

Finally, in the calculations of the wear rate in the T1, T2, and T3 surface hardening conditions, the data obtained showed a direct correlation, both with the hardness data obtained for the three conditions, and with their respective behavior in the values of the hardness rate. reduction in crystallite size. The conditions that presented the highest hardness also presented the lowest wear rates, when compared to the base metal. The same analysis applies when analyzing in relation to the crystallite size reduction rate. Thus, among all the laser surface hardening conditions analyzed, the T3 condition was obtained as the best condition to gain surface mechanical properties, which, among all the

analyzed conditions, combined high values of surface hardness, hardness in the range of depth up to 0.30 mm, surface structure refinement rate, structure refinement rate in the depth range up to 0.30 mm, and low wear rate values.

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Chapter 7

Phase Balancing in Nd:Yag Laser Welding of Preheated UNS S32750 Superduplex Steel

G. S. Crespo* and V. A. Ventrella

Department of Mechanical Engineering, São Paulo State University (UNESP), School of Engineering of Ilha Solteira, Ilha Solteira, SP, Brazil

Abstract

Two UNS S32750 superduplex steel sheets were heated in an oven and subsequently subjected to a Nd:YAG laser welding process in pulsed mode. The first sheet was welded after it was completely cooled in the air (default condition: CP), while the second sheet was welded immediately after its removal from the oven (heated condition: CA). The welding processes in both conditions were carried out in an autogenous manner, at a frequency of 15 Hz, welding speed of 1.7 mm/s, temporal width of 5 ms, peak energy of 2 kW, and without the use of protection. After the welding process, the weld beads were sectioned in their transverse directions, embedded in hot curing Bakelite and subjected to the standard metallographic preparation process, with revelation of the metallographic structure of the weld metals by using Behara II reagent. Macrographs and micrographs were performed under a common optical microscope and, under a scanning electron microscope (SEM), of the weld metals of both conditions, which were used for geometric evaluation (width and depth) of the weld metals and evaluation of the balance volumetric analysis of austenite/ferrite phases, post-welding process. Energy-dispersive spectroscopy (EDS) analyses were also performed on the weld metals under the two conditions, to evaluate the variation, in relation to the base metal, of the three alloying elements (Cr, Ni, and Mo) present in greater amounts. In the geometric analyses, CA weld metal was approximately

* Corresponding author's email: gillian.engmec@gmail.com.

41.8% deeper and 11.2% wider than CP weld metal, indicating a strong influence of sheet heating on laser beam penetration during the welding process. In volumetric phase balancing, the heating of the plate reflected an increase in the average percentage of austenite in the weld metal, from 10.9% to 46.7%, causing a reduction in the austenite phase imbalance, from 39.1% to 3.3%. Finally, in relation to the alloying elements in the weld metals, the CA sample presented a higher Cr value, while the CP sample presented higher Mo and Ni values when compared to each other. In relation to the base metal, both weld metals showed higher percentages for the three alloying elements, in line with the expected behavior for each condition, according to the phase balance data obtained.

Keywords: Heating, UNS S32750, Laser, Nd:YAG

Introduction

The world is experiencing, more and more, drastic effects resulting from climate change arising from industrial processes and from the consumption of fossil fuels. According to the World Meteorological Organization (WMO), in a report published in 2018, from the beginning of the world industrialization process to the present day, the temperature of the planet has risen by almost 1°C. According to the same source, 20 of the warmest years were recorded in the last 22 years, of which the top four were 2015, 2016, 2017, and 2018, according to the WMO (Stylianou 2018).

Much of this impact stems from the use of fossil fuels. The United States (US) and China lead by far in the emission ranking, together accounting for more than 40% of global emissions, according to 2017 data from the Dutch Assessment Agency and the European Commission's Joint Research Center. The US even abandoned the Paris Agreement in favor of a pro-fossil fuels agenda, claiming a policy of not harming American companies and workers (Stylianou 2018).

Attempts to change this calamitous scenario have been underway for some time, with the search for cleaner energy sources, and less aggressive to the environment, an increasingly growing focus in many researches. The bet on the automobile sector, based on electric energy, has also undergone great expansion and investment by a large part of the industrial sector and corporate and political leaders, initiating an era of "electrification" of the automobile industry (Ribeiro 2019).

In Europe, the environmental issue makes governments encourage the manufacture of cars powered by electricity. In France, alcohol, gasoline, and diesel cars will no longer be sold from 2040 onwards. The measure is

focused on respecting the parameters established in the International Paris Agreement against climate change. The UK government has also announced that it intends to end the sale of new cars and vans that use petrol and diesel by 2040. This action aims to reduce carbon dioxide rates on roads as much as possible. Concern about the environment has even reached China, which forced the government to determine that, in 2019, 10% of car sales must include electric or hybrid models. In the following year, this rate will be 12% (Pires 2018).

However, as much as there is a change in the scenario, it will take years for a considerable decrease in the use of fossil fuels in more common sectors. In some consumer markets, such as Brazil, the cost of technology and the little infrastructure currently installed push this change even further into the future. Extending the thinking to other sectors, this estimate is pushed even further. A small boost from other sectors is enough to keep demand for oil and other fossil fuels on an upward trajectory: the use of oil to produce petrochemicals is the biggest source of growth, closely followed by increased consumption by trucks and by aviation and maritime sectors (Days 2018).

Although a change in the use of the current predominant energy matrix is necessary, and this change is taking its first steps, it will still take a long time for this change to be in stride, persevering for a long time in the scenario of exploration and extraction of energy sources, i.e., fossil energy.

Fossil energy reserves are finite. The act of exploitation of these products becomes increasingly challenging, bringing with it severe conditions to the machinery used in its exploitation, regarding environments with extreme demands of mechanical effort and severely chemically aggressive (corrosive). It was precisely to serve this field of the petrochemical industry, as well as other industrial sectors (e.g., pulp and paper, nuclear, chemical, food), that the duplex and superduplex stainless steel families were developed (Franzini 2016; Senatore, Finzetto and Perea 2007; Garcia 2012; Canderl 2016; Oliveira Junior 2013).

These steels consist of a binary structural matrix composed of ferrite and austenite, with an approximate balance of 50% for each phase. This balancing of the binary matrix gives these steels high levels of mechanical strength due to the combination of the high elongation values of austenite and the high yield strength of ferrite (Senatore, Finzetto and Perea 2007). The binary matrix is also responsible for the high levels of corrosion resistance presented by these steels, this resistance refers to pitting corrosion, or simply PRE, responsible for determining the degree of opposition that the material offers to this type of corrosion. In order to be considered a duplex steel, this value must be greater than 20, while for superduplex stainless steel this value must be greater than 40, in severely aggressive environments (Oliveira Junior 2013).

However, when it is necessary to use welding processes in superduplex steels, there may be an unbalance, to a lesser or greater extent, of this binary matrix constituting the material, resulting in parameter losses, either in the mechanical scope or in the scope of resistance to severely chemically aggressive media.

Aiming at the search for parameters that reduce the occurrence of this unbalance, several studies, using different welding techniques, sometimes even being combined with other procedures (Crespo, Videira and Ventrella 2020), have been conducted in different duplex and superduplex steels, in the search for conditions that allow the least possible loss of properties, in relation to the basic characteristics of the material. Thus, the general objective of the developed study focused on the analysis of the phase balance of a superduplex stainless steel (UNS S32750) submitted to a heating process in an oven, followed by a Nd:YAG laser welding process, in the condition, pulsed, performed in “bead on plate” mode.

Within this scope, the specific objectives of the work contemplated the macrographic analysis, from the perspective of qualitative quality of the weld metal, investigating the presence of pores, cracks, or voids; macrographic analysis of the weld metal geometric optics, evaluating the weld metal width and depth configurations; micrographic analysis, by means of scanning electron microscopy (SEM), investigating the volumetric balance of ferrite/austenite phases of weld metals; chemical analysis, using energy-dispersive spectroscopy (EDS), evaluating the percentage variation by weight, in the weld metals, of the three alloying elements in higher concentration in the base metal (Cr, Mo, and Ni).

Methods

The base metal used in the study was superduplex stainless steel UNS S32750, in the form of plates measuring 3 mm × 24 mm × 50 mm (thickness × width × length). The chemical composition of UNS S32750 superduplex stainless steel is shown in Table 7.1.

Table 7.1 Chemical composition of the base metal (% by weight)

C	Si	Mn	P	S	Cr	Mo	Ni	Cu	N
0.018	0.29	0.63	0.02	<0.0003	25.61	3.84	6.97	0.15	0.269

Determination of the Cooling Curve of the Heated Metal

In the first stage of the study, the cooling process of the base metal at room temperature was evaluated after the heating procedure to be used, prior to the execution of the intended laser welding process. The evaluation started with the creation of a small hole in the upper central part of a base metal plate (Figure 7.1), for insertion of a thermocouple, after removing the plate from the heating oven.

The base metal heating process was carried out in an oven at a temperature of 700°C, keeping the metal inside the oven for a sufficient time to reach thermal equilibrium with the oven temperature. After reaching thermal equilibrium, the metal was removed from the oven and the temperature of the plate was measured over time to determine its cooling curve. Temperature measurements were performed using a multimeter in conjunction with a “K”-type thermocouple (chromel and alumel). Measurements were taken at regular intervals of 15 s and obtained in millivolts (mV), later converted to °C, using the conversion table provided by the thermocouple manufacturer. All measurements were performed in an environment free from air currents, or any other type of influence, on the metal cooling process. Figure 7.2 shows the plot of data, in °C, acquired in the first 15 min (900 s) of reading the base metal cooling temperatures, after its removal from the furnace.

Finally, the evaluation of the transport procedure of the part was carried out, from the place of the oven used for the heating process to the place of the laser welding process. For this, the piece was transported, protected by refractory bricks, from the heating place to the welding place. The temperature of the part, measured after transport between locations, showed values consistent with the measurement performed immediately after removal from the oven. Transport time was calculated between 1.5 min (90 s) and 2 min (120 s). Analyzing the graph in Figure 7.2, and considering the transport parameters of the part from the oven location to the welding location, it was concluded that

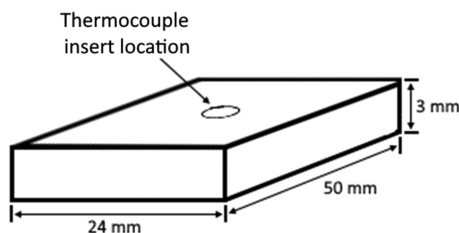


Figure 7.1 Plate used to determine the cooling curve of the base metal after heating in an oven.

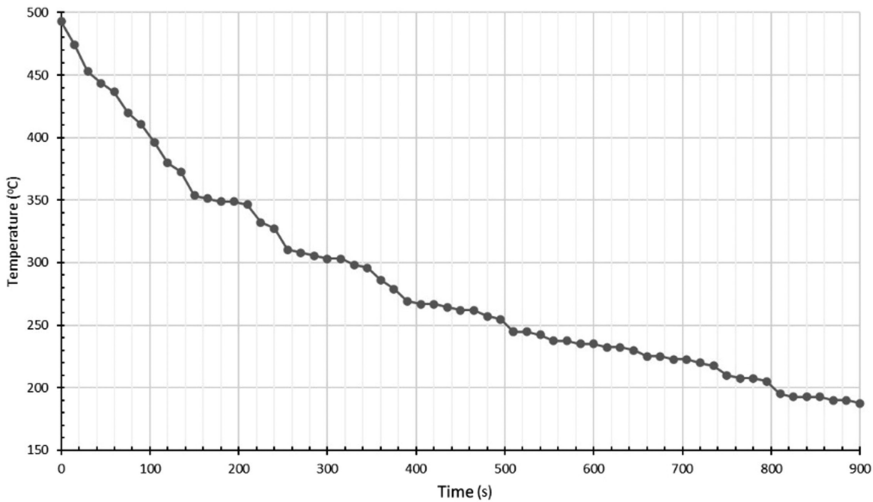


Figure 7.2 Graph of base metal cooling measurements for the first 900 s.

Table 7.2 Operating parameters of the Nd:YAG laser system, model UW 150 A

Wave length	1064 nm
Maximum laser power	150 W
Peak power	7 kW
Maximum pulse energy	80 J
Pulse width	0,5 ms–30 ms
Frequency	1 Hz–300 Hz
Positioning system	Red laser and CCD camera

the welding process took place with the plate at an approximate temperature of 400°C.

Laser Welding Process and Metallographic Preparation

In the second stage of the study, the base metal was welded. The base metal was welded in two conditions: CP (standard condition, with the base metal heated in an oven to 700°C and cooled in still air before the welding process) and CA (heated condition, with the base metal heated in an oven to 700°C and welded at an approximate temperature of 400°C). The laser welding process

Table 7.3 Standard (CP) and heated (CA) conditions welding parameters

Specimen	CP	CA
Shielding gas	Sem gás	Sem gás
Welding speed	1.7 mm/s	1.7 mm/s
Frequency	15.0 Hz	15.0 Hz
Power	2.0 kW	2.0 kW
Temporal width	5.0 ms	5.0 ms
Pulse energy	10.0 J	10.0 J
Specimen temperature	26.0°C	~400°C

used a Nd:YAG laser system, model UW 150 A, from United Winners, in the pulsed condition. Welding was performed in the “bead on plate” model, a model where the weld bead is performed on the flat surface of a single plate. The two welding conditions were performed without the use of shielding gas, in order to avoid the cooling of the heated part. Tables 7.2 and 7.3 show, respectively, the operational parameters of the laser system used and the welding conditions performed.

After carrying out the welding process, the welded plates were sectioned and hot embedded in Bakelite, generating a sample for each welding condition performed. Then, each sample underwent a conventional metallographic preparation process. The metallographically prepared surfaces were then etched with Behara II reagent (100 ml of H₂O + 20 ml of HCl + 0.4 g of potassium metabisulfite), with an etching time of 20 s, to reveal the structure of the weld metals obtained. After the metallographic preparation and structure revelation process, the weld metals of each condition were subjected to analysis: macrographic (geometry and qualitative quality of the weld metal), micrographic (volumetric balancing of ferrite/austenite phases in the weld metal), and chemical (variation of alloying elements in the weld metal).

Macrographic Analysis

The macrographic analyses (geometry and qualitative quality of the weld metal) were carried out through the computational measurement of width and depth (Figure 7.3), from macrographic images obtained in a common optical microscope, of the weld metals in each welding condition performed.

The objective of this step was to investigate the geometric variation of the weld metal, as well as the evaluation of the qualitative quality of the weld metal, looking for the presence of pores, cracks, and voids.

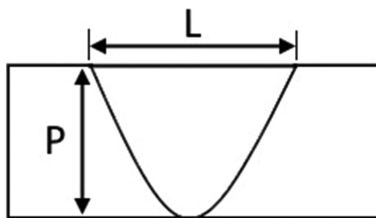


Figure 7.3 Geometric measurements performed on the weld metal of each welding condition: L (width) and P (depth).

Micrographic Analysis

The micrographic analyses (volumetric balancing of ferrite/austenite phases in the weld metal) were conducted through the computational calculation of the percentage of austenite present in the weld metal, from micrographic images of the weld metals obtained in a SEM, brand JEOL, model JSM-T330A.

The computational calculation of the percentage of each phase in the weld metal was performed using the “Image J” program, public domain image editing software based on the Java platform. The objective of this step was to evaluate the volumetric balance of the ferrite and austenite phases in the weld metals, when compared with the percentages (approximately 50% each) present in the base metal as received from the factory.

Chemical Analysis

Chemical analysis was carried out by conducting EDS performed on a JEOL scanning electron microscope, model JSM-T330A. The EDS analysis was performed by performing four analysis lines on the weld metals of each welding condition, with three horizontal lines (one located close to the bead surface, one located in the center of the bead, and one located close to the bead surface at the root of the cord) and a vertical line (starting near the surface of the cord and extending to the region close to the root of the cord). In all cases, the lines were started and ended within the weld metal region and away from the transition regions between weld metal and base metal, ensuring that there was no interference of results due to the composition of the base metal.

The objective of this step was to investigate the structure of the weld metals, regarding the variation of the percentage by weight of the three alloying elements (Cr, Mo, and Ni) in greater presence in the composition of the base metal as received from the factory, verifying the occurrence of any significant

change in the balance of such elements after the welding process, for the CP and CA conditions.

Results and Discussion

Macrographic Analysis

Figures 7.4 and 7.5 show, respectively, the macrographs of the CP and CA samples.

The CP sample (Figure 7.4) shows the common result obtained when laser welding duplex and superduplex steels. The weld metal presents islands of ferrite, isolated by branches of austenite with an origin of growth from the edges of the weld metal, walking toward the interface of the base metal with the weld metal, to the center of the bead. In this condition, it is also possible to distinguish between the regions coming from different welding pulses, such regions being separated by lines of the same contour as the weld bead.

As for the CA sample (Figure 7.5), we have a completely different behavior. We still have branches of austenite similar to the CP sample, but now we have what appears to be a swirling behavior of the weld metal, creating flux zones and austenite concentration. In the CA sample, it is also no longer possible to make a clear distinction between the zones of the weld metal arising from each pulse of the welding process. Table 7.4 shows the width and depth data obtained for the CP and CA samples.

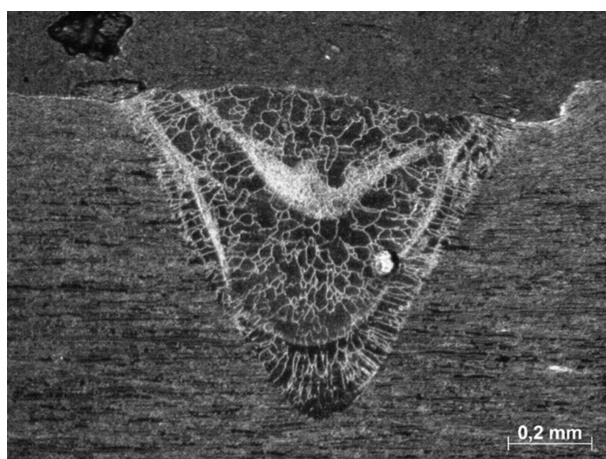


Figure 7.4 Macrograph of the CP sample.

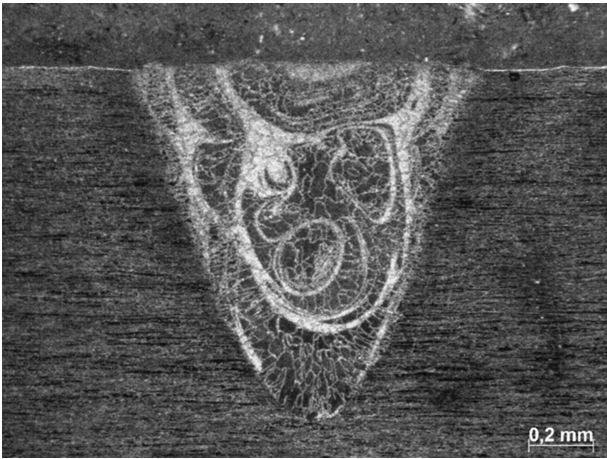


Figure 7.5 Macrograph of the CA sample.

Table 7.4 Width and depth data for CP and CA samples

Welding condition	Width (mm)	Depth (mm)
CP	0.98	0.79
CA	1.09	1.12

By analyzing the data contained in Table 7.6, we infer that the weld metal of the CA sample presented higher geometric values than the weld metal of the CP sample, which allows us to attest that the temperature of the plate directly influenced the geometry obtained by the weld metal, resulting in a wider (11.2%) and deeper (41.8%) bead for the welded sample while heated (CA).

This can be attributed to the fact that when the laser hits the sheet, it has to melt it, layer by layer, to penetrate toward the depth of the sheet. In this way, once the sheet is already heated, the time for heating each layer to the melting point decreases, allowing the laser to penetrate more deeply into the CA sample than into the CP sample, considering the same temperature interval. The same analysis can be applied regarding the width of the weld metal, so that the lateral layers of the weld region of the CA sample reach the melting point primarily to the lateral layers of the weld region of the CP sample, resulting in an increase in width of the weld metal of the CA sample in relation to the CP sample.

Micrographic Analysis

Tables 7.5 and 7.6 show, respectively, the values of the percentages of the austenite and ferrite phases for each image of the CP and CA samples, together

Table 7.5 Values of the percentages of the austenite and ferrite phases for each image of the CP sample, together with the arithmetic mean and standard deviation values for each phase

Image	Austenite (%)	Ferrite (%)
1	12.4	87.6
2	9.8	90.2
3	10.6	89.4
4	10.6	89.4
5	11.9	88.1
6	10.9	89.1
7	10.4	89.6
Mean	10.9	89.1
Standard deviation	0.9	0.9

Table 7.6 Values of the percentages of the austenite and ferrite phases for each image of the CA sample, together with the arithmetic mean and standard deviation values for each phase

Image	Austenite (%)	Ferrite (%)
1	48.5	51.5
2	45.0	55.0
3	45.2	54.8
4	49.2	50.8
5	44.3	55.7
6	43.9	56.1
7	51.2	48.8
Mean	46.7	53.3
Standard deviation	2.9	2.9

with the arithmetic mean and standard deviation values for each phase. Figures 7.6 and 7.7 show, respectively, the set of seven images of the weld metal of the CP and CA samples used to calculate the volumetric balance of the austenite and ferrite phases in the weld metal of each sample.

By analyzing the data in Tables 7.5 and 7.6, we can see that the heating of the plate resulted in an increase in the average percentage of austenite in the weld metal, rising from 10.9% in the CP sample to 46.7% in the CA sample. Considering the standard balance of the base metal, which has approximately 50% of its matrix constituted by the austenite phase, we have a reduction of the unbalance of the austenite phase from 39.1% to 3.3%. The standard deviations

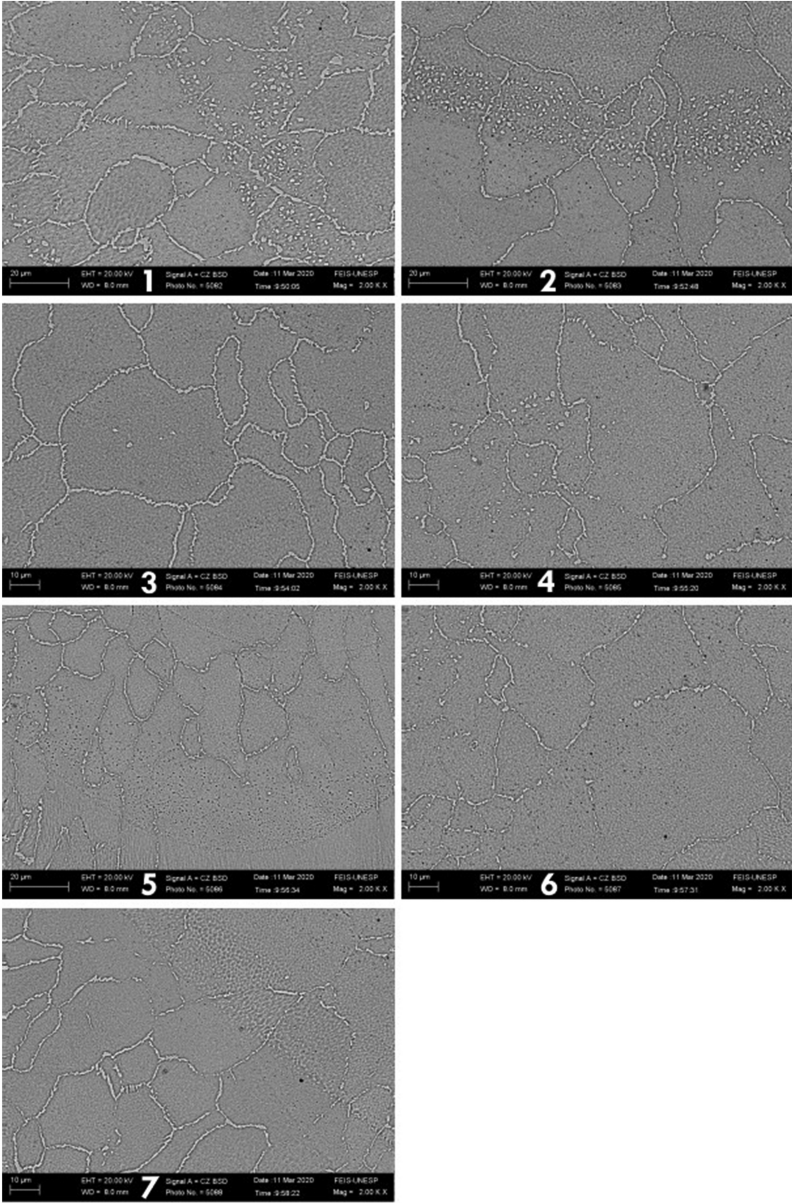


Figure 7.6 Set of seven weld metal images, from the CP sample, used to calculate the volumetric balance of austenite and ferrite phases.

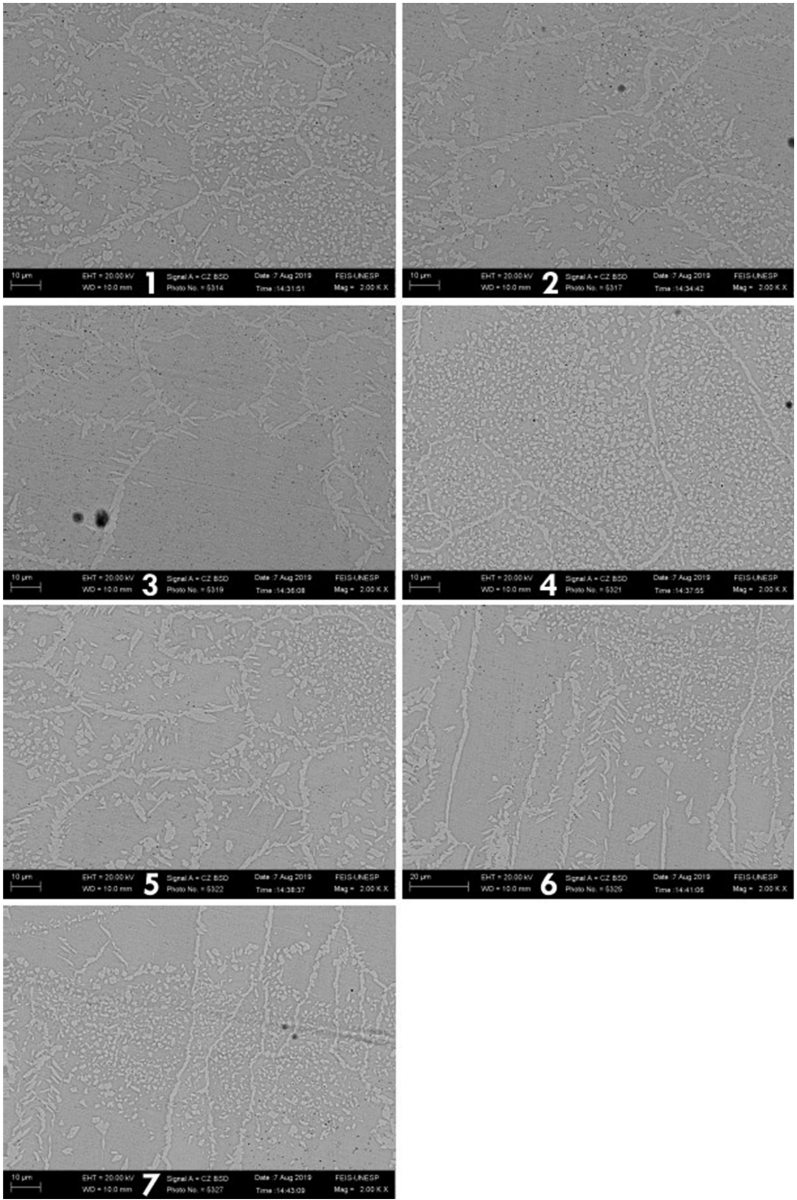


Figure 7.7 Set of seven weld metal images, from the CA sample, used to calculate the volumetric balance of austenite and ferrite phases.

in the analysis of the two welding conditions showed a value below 3.0%, which reflects a low amplitude of variation within the analyzed data.

This increase in the austenite phase in the weld metal of the CA sample can be attributed to the fact that, due to the initial temperature of the plate, at the beginning of the welding process, there was a longer time for stabilization of austenite in the weld metal after conclusion of the process, since, as a result of the increase in time for cooling the molten pool, there was a decrease in the drastic cooling rates commonly present in laser welding processes.

In duplex and superduplex steels, the corrosion resistance aspect presented by these materials is attributed to the austenitic phase. Therefore, with a higher percentage of austenite in the CA sample, we can infer that it will present better corrosion resistance results when compared to the CP sample. In relation to the ferritic phase, the aspect of mechanical resistance presented by these materials is attributed to it. Therefore, with a higher percentage of ferrite in the CP sample, we can infer that it will present better mechanical behavior results when compared to the CA sample.

Chemical Analysis

Figures 7.8 and 7.9 show, respectively, the four lines of energy-dispersive spectroscopy (EDS) analyzed in the weld metals of the CP and CA samples.

Figures 7.10–7.12 show, respectively, the plot of the values of the percentage weights obtained for the elements Cr, Mo, and Ni, for each of the four EDS

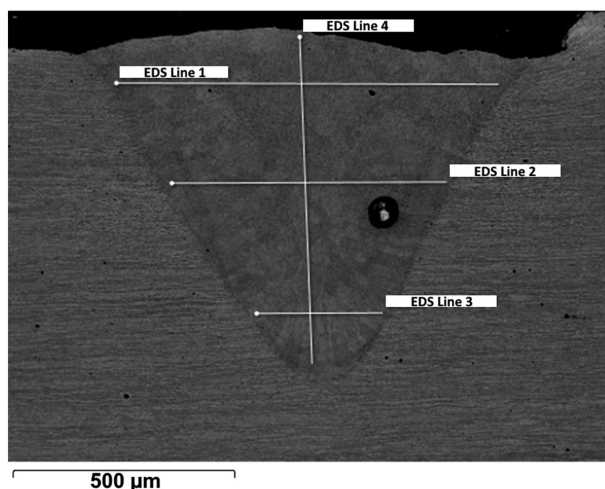


Figure 7.8 Energy-dispersive spectroscopy (EDS) lines analyzed in the CP sample.

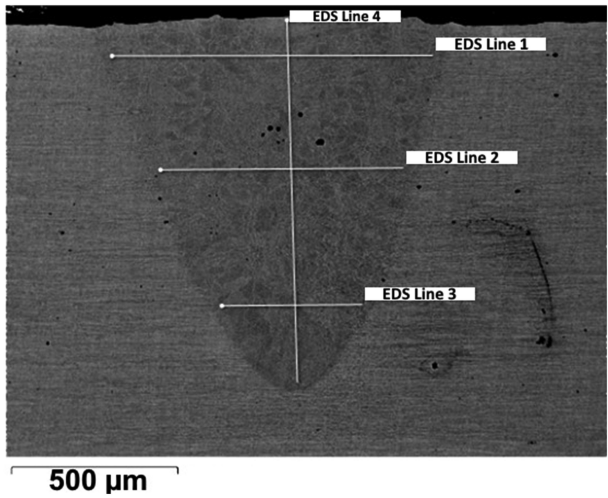


Figure 7.9 Energy-dispersive spectroscopy (EDS) lines analyzed in the CA sample.

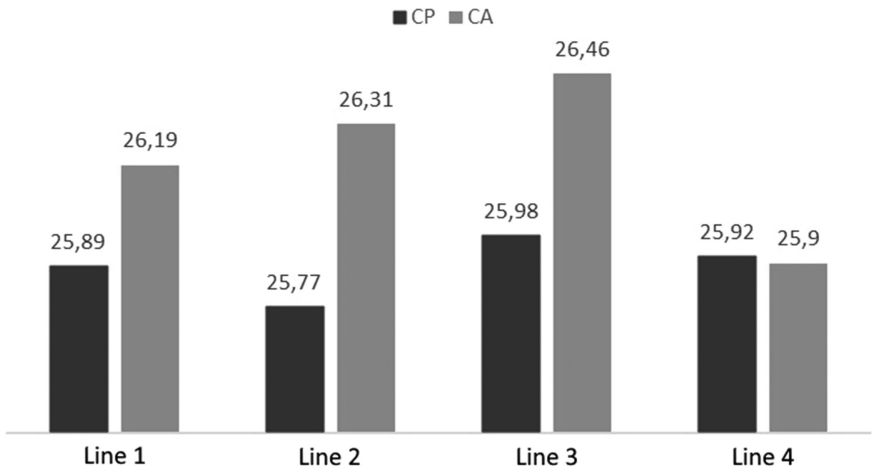


Figure 7.10 Average weight (%) values for the Cr element, for each of the four EDS lines performed on the CP and CA samples.

lines performed in the CP and CA samples. Tables 7.7 and 7.8 show, respectively, the values of the average percentage weight per line and the global average (M.G.) and global average standard deviation (D.P.M.G.) of the average percentage weights of the elements Cr, Mo, and Ni, in the analysis by EDS for the CP and CA samples.

Tables 7.9 and 7.10 show, respectively, the percentage variations for the CP and CA samples between the global average (M.G.) of the weights obtained in

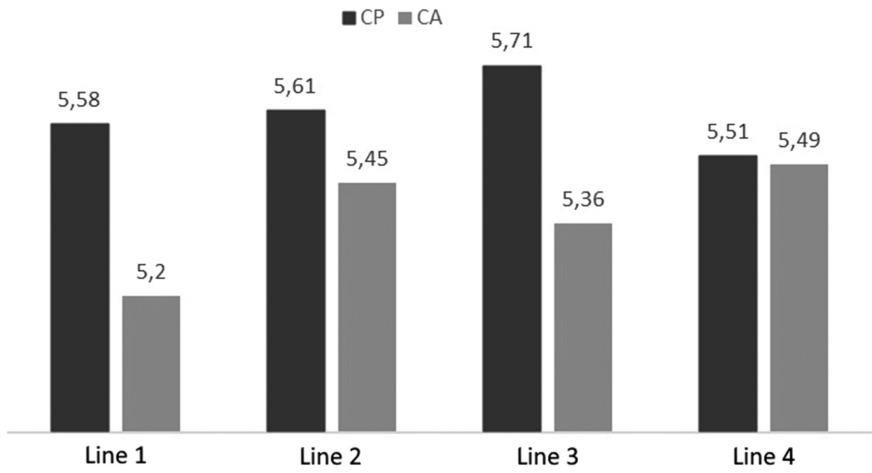


Figure 7.11 Average weight (%) values for the Mo element, for each of the four EDS lines performed on the CP and CA samples.

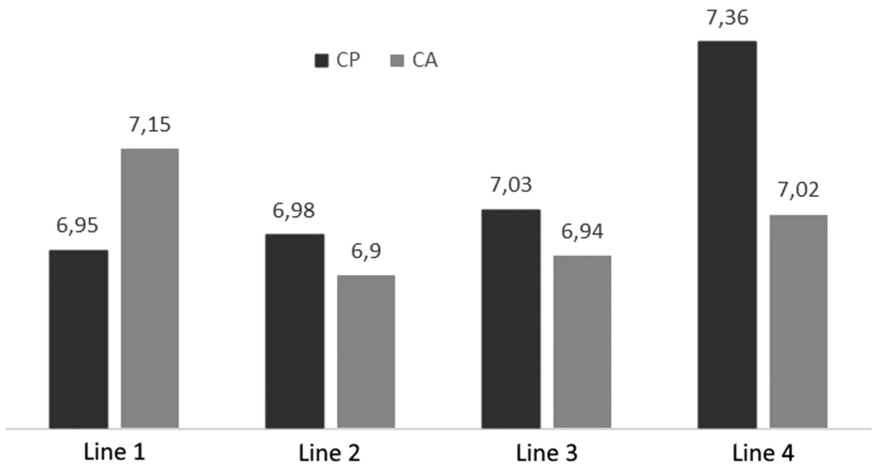


Figure 7.12 Average weight (%) values for the Ni element, for each of the four EDS lines performed on the CP and CA samples.

the EDS analysis for the elements Cr, Mo, and Ni and the global average (M.F), provided by the manufacturer for these same elements.

Analyzing the EDS data, we have that the CA sample presented higher Cr values in the three horizontal lines and, a value close to the vertical line, in relation to the CP sample, resulting in an average value of 26.22% for the CA sample, against an average value of 25.89% for the CP sample. In the

Table 7.7 Values of the average percentage weight per line and global average (M.G.) and global average standard deviation (D.P.M.G.) of the average percentage weights of the elements Cr, Mo, and Ni, in the EDS analysis of the CP sample

Element	Line 1	Line 2	Line 3	Line 4	M.G.	D.P.M.G.
Cr	25.89	25.77	25.98	25.92	25.89	0.09
Mo	5.58	5.61	5.71	5.51	5.60	0.08
Ni	6.95	6.98	7.03	7.36	7.08	0.19

Table 7.8 Values of the average percentage weight per line and global average (M.G.) and global average standard deviation (D.P.M.G.) of the average percentage weights of the elements Cr, Mo, and Ni, in the EDS analysis of the CA sample

Element	Line 1	Line 2	Line 3	Line 4	M.G.	D.P.M.G.
Cr	26.19	26.31	26.46	25.90	26.22	0.24
Mo	5.20	5.45	5.36	5.49	5.38	0.13
Ni	7.15	6.90	6.94	7.02	7.00	0.11

Table 7.9 Percentage variation between the global average (M.G.) of the weights obtained in the EDS analysis and the global average (M.F.) provided by the manufacturer, for the weld metal of the CP sample

Element	M.G.	M.F.	Variation (%)
Cr	25.89	25.61	1.09
Mo	5.60	3.84	45.90
Ni	7.08	6.97	1.58

Table 7.10 Percentage variation between the global average (M.G.) of the weights obtained in the EDS analysis and the global average (M.F.) provided by the manufacturer for the weld metal of the CA sample

Element	M.G.	M.F.	Variation (%)
Cr	26.22	25.61	2.36
Mo	5.38	3.84	39.97
Ni	7.00	6.97	0.47

comparison with the base metal, the variation reached 2.36%, against 1.09% of the CP sample. Cr is an alloying element that is attributed as one of its functions, with an increase in corrosion resistance. Thus, we can infer that the CA sample will present better corrosion resistance results when compared to the CP sample. This analysis is suitable for volumetric phase balance analysis, since it is expected that the CA sample presents better corrosion resistance results due to the greater presence of austenite in its balance.

In relation to the Mo element data, the CA sample presented lower values of Mo in the three horizontal lines and a close value in the vertical line, in relation to the CP sample, resulting in an average value of 5.38% for the CA sample, against an average value of 5.60% for the CP sample. In the comparison with the base metal, the variation reached 39.97%, against 45.90% of the CP sample. Mo is an alloying element to which an increase in mechanical strength is attributed as one of its functions. Thus, we can infer that the CP sample will present better mechanical strength results when compared to the CA sample. This analysis is suitable for volumetric phase balance analysis, since it is expected that the CP sample presents better mechanical results due to the greater presence of ferrite in its balance.

Finally, in relation to the Ni element data, the CA sample presented a higher value only in the first horizontal line, located close to the surface of the weld bead executed, presenting in all the other three lines, values lower than the CP sample, resulting in an average value of 7.00% for the CA sample, against an average value of 7.08% for the CP sample. In the comparison with the base metal, the variation reached 0.47%, against 1.58% of the CP sample. Ni is an alloying element that is attributed, as one of its functions, an increase in corrosion resistance, in addition to being a categorized gammagenic element, which helps in the formation and stabilization of austenite. Although a higher Ni value was expected in the CA sample, in relation to the CP sample, due to the results visualized in the performed phase balancing, the CA sample still presented values higher than the base metal, a result consistent with the other factors involved in the process.

Furthermore, although the CP sample showed a higher percentage of Ni in the weld metal, the variation of the austenite phase in the weld metal also depends on other factors involved in the process, such as the residence time at high temperatures, which is found directly linked to the cooling rates experienced by the weld metal, and that in the CP sample had higher speed, compared to the CA sample, due to the base metal being at room temperature during the welding process. Thus, although it presents a greater amount of Ni, the CP sample ends up showing a lower amount of austenite phase than that seen in the CA sample, which has a slower cooling of the weld metal, as the base metal has been welded in a state heated.

Conclusion

The CA sample, where heating was performed, prior to the welding process, showed wider (11.2%) and deeper (41.8%) weld beads when compared to the CP sample, which received the same welding process, but without the presence of prior heating of the base metal. This variation in values can be attributed to the fact that, once the base metal is heated, the laser beam makes it easier to reach the melting temperature of the metal, resulting in greater molten volumes in the same time interval.

Regarding the qualitative behavior, the weld metal of the CP sample showed islands of ferrite, isolated by branches of austenite toward the interface of the base metal with the weld metal, up to the center of the bead, in addition to a distinction between the regions coming from different welding pulses. As for the CA sample, we observed austenite ramifications similar to those of the CP sample, but with a predominance of a swirling behavior of the weld metal, creating flux zones and austenite concentration.

The CA sample showed an increase in the average percentage of austenite in the weld metal, rising from 10.9% in the CP sample to 46.7% in the CA sample, which resulted in a reduction of the austenite phase imbalance of 39.1% to 3.3%. With a higher presence of austenite phase, the CA sample should present better corrosion resistance results compared to the CP sample.

Finally, in relation to the analysis of alloying elements in the weld metals, the CA sample presented the highest Cr value, while the CP sample presented the highest Mo and Ni values. The results are consistent with the results observed in the phase balancing of the samples, with the expected condition of presenting better corrosion resistance (CA) results, with Cr and Ni values higher than the base metal, while the expected condition of presenting better results mechanical (CP), with higher values of Cr and Mo in relation to the base metal. Although the Ni values were higher in the CP sample, the presence of austenite in the phase balancing is also a factor dependent on the thermal cycle experienced by the weld metal, so that, although with a lower amount of Ni, the CA sample ends up showing higher austenite values in the weld metal compared to the CP sample.

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Chapter 8

Correction of the Phase Balance on Nd:YAG Pulsed Laser Welded Super Duplex Stainless Steel by Nickel and Cobalt Electroplating

**E. J. da Cruz Junior¹, F. M. F. A. Varasquim¹,
B. B. Seloto², V. A. Ventrella², A. G. Settimi³,
C. Gennari¹, A. Zambon³ and I. Calliari³**

¹São Paulo Federal Institute of Education, Science and Technology, Campus Itapetininga, Av. João Olímpio de Oliveira 1561, 18202-000, Itapetininga, SP, Brazil

²Mechanical Engineering Department, Campus of Ilha Solteira, UNESP, São Paulo State University, Av. Brasil 56, 15385-000, Ilha Solteira, SP, Brazil

³Department of Industrial Engineering, University of Padua, Via Gradenigo, 6/a – 35131, Padova, Italy

Abstract

With a volume fraction of austenite and ferrite that is roughly equal, super duplex stainless steels (SDSS) have good mechanical characteristics and a high level of corrosion resistance. Achieving an equilibrium microstructure in the fusion zone, and ensuring its qualities and uses, is the difficulty in SDSS laser welding. In general, when SDSS is welded with a technique inherently associated with high cooling rates, a mostly ferritic microstructure results. To maintain the alloy's mechanical properties and corrosion resistance, ferrite and austenite must remain in roughly equal quantities. To balance the microstructure, the addition of austenite stabilizing elements or heat treatments is indicated. The addition of alloying materials is challenging due to the Nd:YAG pulsed laser welding properties. An approach is to use metal electroplating to prepare the surfaces of the mechanical SDSS components that will be welded, therefore promoting the phase balance in the fusion zone. This work examined the use of metals electroplating techniques to encourage the enrichment of nickel

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and cobalt in the bead of UNS S32750 to be welded by pulsed Nd:YAG laser source and, as a result, prevent the phase imbalance. This was done because phase balance is important to the SDSS's characteristics and applications. Prior to butt-welding, two UNS S32750 SDSS sheets with a pulsed Nd:YAG laser, a nickel and cobalt layer, were electrodeposited onto the edge of one of them with a Watts bath procedure. By adjusting the deposition periods, different thicknesses of nickel and cobalt were utilized, and the microstructure of the weld bead was assessed. As the volume fractions of austenite clearly increased, the results demonstrated that the nickel and cobalt added in the fusion zone were effective in resolving the unwanted unbalanced microstructure that resulted from autogenous laser welding of the SDSS.

Keywords: Nickel; Cobalt; Laser welding; Electroplating; Volume fraction; Super duplex stainless steel

Introduction

Super duplex stainless steels (SDSSs) exhibit a dual-phase microstructure of nearly equal ferrite and austenite proportions (Zhang et al., 2017). They offer several advantages for single-phase grades, including good mechanical properties and corrosion resistance in specific environments (Ramkumar et al., 2014). SDSS is frequently used when mechanical and corrosion resistance is needed (Saravanan et al., 2017).

The SDSS is well protected against pitting attacks in chloride-containing solutions, thanks to the significant levels of Cr, Mo, and N present. Equation 8.1 shows that the pitting resistance number (PREN) can be used to rank pitting susceptibility in Duplex stainless steel (DSS) (Kang et al., 2013)

$$\text{PREN} = \text{wt.\%Cr} + 3.3 \text{ wt.\%Mo} + 16 \text{ wt.\%N} \quad \text{Equation 8.1}$$

The difference between DSS and SDSS beyond the concentration of alloying elements is their PREN. DSS has PREN between 30 and 40, while SDSS has PREN above 40.

Despite the broad advantages of SDSS, after undergoing the welding process, SDSS presents an unbalanced microstructure with high content of ferrite (Arun et al., 2019, Ramkumar et al., 2017). To maintain the alloy's mechanical and corrosion resistance qualities, ferrite and austenite must remain in roughly equal quantities (Ouali et al., 2019). Unfortunately, SDSSs typically exhibit an unbalanced microstructure after high power density welding processes, with ferritic phase predominating, which reduces the hardness, local pitting, and

corrosion resistance of SDSS (Tan et al., 2012). Another challenge is the possibility of harmful secondary phases precipitating in the temperature range between 700°C and 950°C, depending on the welding technique and other parameters. The mechanical characteristics and corrosion resistance of the SDSS will be impacted by these phases (Pezato et al., 2018). High energy input during the welding process can result in deformed weld beads, a wide heat-affected zone (HAZ), and poor corrosion resistance (Mei et al. 2020).

The development of new DSS grades, such as SDSS, demands the use of more appropriate welding techniques. The Nd:YAG pulsed laser welding process has some benefits over traditional methods, including process precision, rapid cycle times, and minimal heat input. However, the technique's typical low heat input and rapid cooling rates favor the development of ferrite, leading to an imbalanced microstructure (Hu et al., 2018).

The Nd:YAG laser is inserted among the solid-state lasers. The term YAG means Yttrium-Aluminum-Garnet ($Y_3Al_5O_{12}$) and is a garnet group material commonly used as the basis for various solid-state lasers. The triple-ionized neodymium dopant replaces a small fraction of yttrium ions in the host crystal structure as their sizes are similar. It is the neodymium ions that provide the laser activity in the crystal.

In conduction laser welding, the base material is heated by the laser beam to its melting temperature without creating vaporization (Benter, Petring and Poprawe, 2015). In keyhole welding, the laser intensity within the focusing position is high enough to generate a high evaporation rate. With this, the recoil pressure deconstructs the melting pool allowing the laser beam to penetrate deeper into the material (Steen and Mazumder, 2010). In both welding techniques, they can be performed with laser in pulsed or continuous mode. The primary factors that determine whether conduction or penetration welding will occur during pulsed laser welding are the length of the laser pulse and the intensity of the laser beam on the component to be welded.

In a study, Muthupandi et al. (2005) examined the effects of nickel and nitrogen addition on the microstructure and mechanical characteristics of metals used in DSS welds. They used an UNS 31803 DSS as the base metal. First, the samples were welded with Gas Tungsten Arc Welding (GTAW) using a nickel-enriched filler and a shielding gas mixture containing enriching the welds with nickel or nitrogen. The samples were re-melted using power laser beam. They concluded that the addition of nickel produced welds with acceptable ferrite–austenite balance. Similar research was conducted by Migiakis and Papadimitriou (2009), however it focused on plasma welding.

The influence of nickel on the toughness of lean DSS weld metal produced by submerged arc welding was examined by Pilhagen and Sandström (2014). An UNS 32101 DSS was utilized as the base metal. They came to the

conclusion that adding nickel produced weldments with the right amount of ferrite. The toughness and ductility were greatly improved.

The impact of nickel powder addition and post-weld heat treatment on the metallurgical and mechanical properties of the welded UNS S32304 DSS was investigated by Tahaei et al. (2016). They came to the conclusion that the mechanical properties of the weld joints might be enhanced by adding nickel powder both before and after the welding operation. In comparison to the addition of nickel powder, PWHT had the best results in restoring an equal percentage of ferrite and austenite.

The Nd:YAG pulsed laser welding characteristics make the addition of alloying elements difficult (Da Cruz Junior et al., 2019). Placing a thin nickel foil between the structural components before welding is an alternate technique that has been discussed in the literature in order to encourage nickel enrichment in the fusion zone (Da Cruz Junior et al., 2021). However, this method has several drawbacks, such as its use in complex-form welding (Da Cruz Junior et al., 2022).

The Watts bath is the most used nickel and cobalt electroplating bath because it is regarded as a cost-effective and effective method. Nickel electroplating is usually used for preparing coating on many metallic materials, improving their corrosion resistance [Salcedo et al., 2016; Wang et al., 2018]. As mentioned, there are several studies concerning the addition of nickel to obtain a balanced microstructure. Although cobalt is an austenitizing element (Lippold and Kotecki, 2005; Helis et al., 2009), its application is little studied to balance the SDSS microstructure during a welding process. Cobalt is used to improve wear resistance, corrosion resistance, and heat resistance in cobalt-base alloys and nickel-base superalloys. Cobalt reduces the grain boundary carbides precipitation, improving corrosion resistance (Neville and Hodgkiess, 1998).

There is limited research on using Watts bath to prepare the surfaces of SDSS mechanical elements that will be welded and thereby encourage the phase balance in the fusion zone. The use of metal electroplating techniques associated with laser welding process has been shown to be a very promising solution for the phase balance correction on SDSS welding (Da Cruz et al., 2022).

Given the significance of maintaining phase balance to preserve the characteristics of SDSSs and, consequently, their applications, as well as the suitability of metal electroplating techniques for forming an electrodeposited cobalt and nickel layer and the paucity of studies regarding cobalt's effect as an austenite-promoting element in SDSS welding this work studied the use of metals electroplating techniques as a way to promote the enrichment of nickel and cobalt in the bead of UNS S32750 to be welded by pulsed Nd:YAG laser source and, consequently correct the phase balance on the resulting microstructure.

Materials and Methods

An UNS S32750 SDSS base material in 1.5-mm-thick sheets was the material employed. Table 8.1 provides the chemical composition, while Figure 8.1 displays the microstructure.

Part of the samples' surface was immersed in the Watts bath, facing the joint regions, in order to obtain an electrolytically deposited nickel and cobalt thin film with a specified thickness. For the nickel deposition, the chemical composition of the bath's solutions was 9 g of $\text{NiCl}_2 \cdot 6\text{H}_2\text{O}$, 52.6 g of $\text{NiSO}_4 \cdot 6\text{H}_2\text{O}$, 7.6 g of H_3BO_3 in 200 ml of H_2O . Temperature and pH were both 4 and 50°C , and the current density was 5 A/dm^2 (Rose, 2014). For the cobalt deposition, the chemical composition of the bath's solutions was 31.2 g of $\text{CoSO}_4 \cdot 7\text{H}_2\text{O}$, 2.0 g of NaCl , 7.6 g of H_3BO_3 in 100 ml of H_2O . Temperature and pH were both 2 and 30°C , and the current density was 3.7 A/dm^2 (Panda, 2008). Thereafter the samples were welded in pairs, with only one part having the electrodeposited layer on it, as presented in Figure 8.2.

The used Nd:YAG pulsed laser system (UW-150A) had a 150 W maximum power and a beam spot diameter of 0.2 mm. The weld that achieves a regular surface, without porosity, and with a deep weld pool greater than 50% of the sheet thickness was analyzed to determine the welding parameters. The focal

Table 8.1 Composition of UNS S32750 (wt%)

Cr	Ni	Mo	Mn	Si	N	Cu	P	C
25.61	6.97	3.84	0.63	0.29	0.27	0.15	0.02	0.018

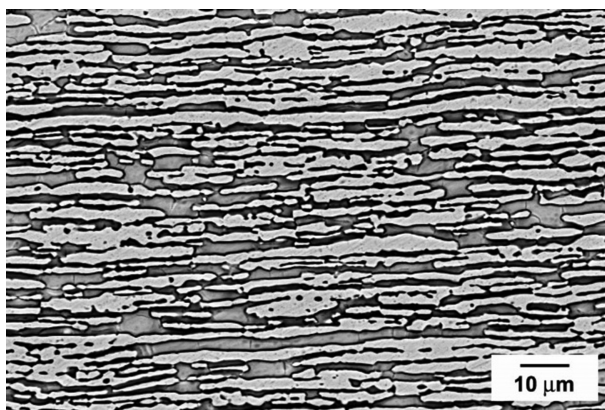


Figure 8.1 SEM micrograph of UNS S32750 SDSS.

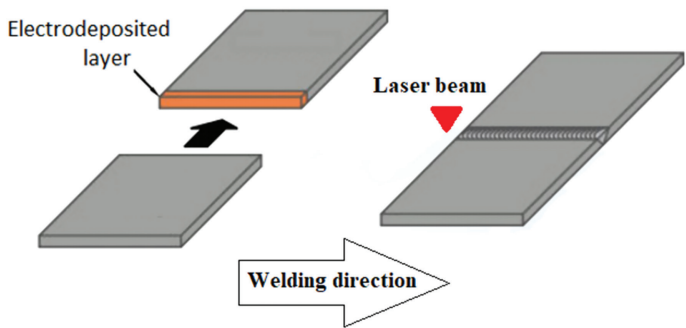


Figure 8.2 Representation of weld joint configuration.

Table 8.2 Welding conditions and the thickness of the electrodeposited layer

Condition	Electrodeposited metal	Layer thickness [μm]	Welding parameters
AW	None	0	Peak power: 2.0 kW Pulse duration: 5 ms Frequency: 9 Hz Welding speed: 1 mm/s Shielding gas: Argon Flow rate: 20 l/min
NiW	Nickel	9.1 ± 1.3	
CoW	Cobalt	35.4 ± 1.1	

point of the laser beam, operated in keyhole mode, was the sample surface and the pulse energy was set at 10 J. Three conditions were evaluated: autogenous welding (AW), welding with nickel addition (NiW), and welding with cobalt addition (CoW), all with the same welding parameters. Table 8.2 lists the welding conditions and the resulting thickness of the electrodeposited layer for the conditions.

Following the welding procedure, the samples were cut transversely with regard to the direction of the weld and then received a typical metallographic preparation. The microstructure was revealed using Beraha’s reagent. The austenite/ferrite volume fraction was evaluated using SEM-BSE images and ImageJ software.

Vickers microhardness was evaluated perpendicular to the welding direction across the entire width of the weldments, with a test load of 10 gf for 10 s and intervals of 0.08 mm respecting the ASTM-E384 standard. Energy-dispersive X-ray spectroscopy (EDS) analyses were performed to determine the amount (wt.%) of nickel and cobalt on the electrodeposited layers and weld beads.

Results and Discussion

The optical micrograph and EDS diagram of the metal electrodeposited layer for the samples NiW and CoW are presented in Figure 8.3. About 97% of the nickel electrodeposited layer's composition was nickel, with the remaining 3% made up of Cr and Fe from the substrate. The carbon peak has been found because the sample is embedded in bakelite. The composition of the cobalt electrodeposited layer was approximately 95% Co and the remaining 5% Fe and Cr, probably also from the substrate. The nickel and cobalt electrodeposited layers displayed uniform thickness and good adherence to the surface of the base metal. The distribution of nickel and cobalt along the weld bead would not be consistent if the thickness weren't constant, resulting in variable austenite/ferrite proportions. Porosity, which if present, could lead to welding problems, wasn't present (Mohammed et al., 2017).

Figure 8.4 depicts the microstructures of the weld bead under all conditions. The microstructure of the AW sample is severely unbalanced (Figure 8.4 a), exhibiting a high proportion of ferrite polygonal grains with austenite decorating their boundaries, which is a common outcome of Nd:YAG pulsed laser welding performed on SDSSs. Austenite is formed through a solid-state diffusion mechanism during DSS welding. Heat input and cooling rates

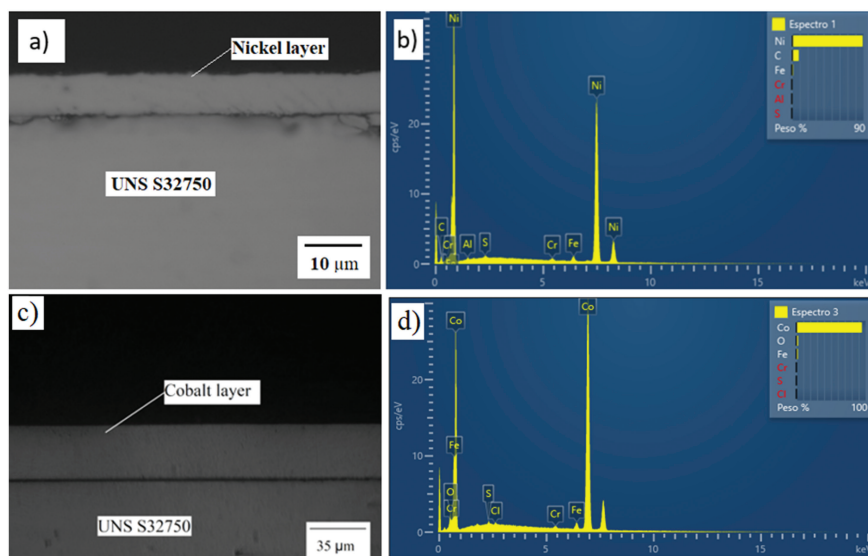


Figure 8.3 (a) Optical images of nickel layer – NiW, (b) EDS nickel layer – NiW, (c) optical micrographs of cobalt layer – CoW, and (d) EDS cobalt layer – CoW.

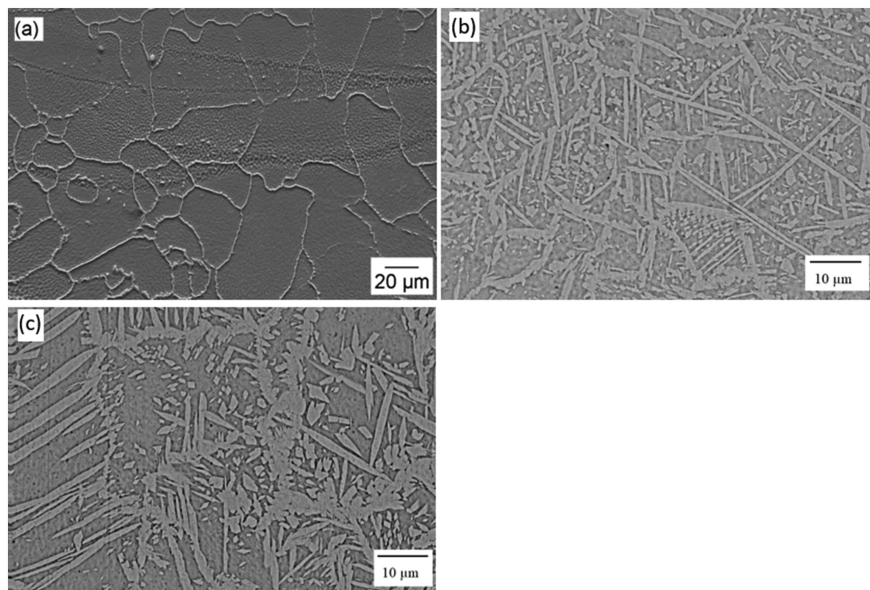


Figure 8.4 SEM images of weld bead for (a) autogenous welding, (b) NiW, and (c) CoW.

experienced by the weld bead during the thermal cycle will have an impact on austenite development. High cooling rates, a particular characteristic of laser welding, don't provide the ferrite-to-austenite transformation enough time to finish, resulting in a primarily ferritic final microstructure (Mohammed et al., 2017). The mechanical characteristics and corrosion resistance of SDSS welds may deteriorate due to the ferritic phase's predominance (Tomashchuk et al., 2015).

Both the addition of nickel and cobalt has affected the amount of austenite and its morphology. The presence of the nickel and cobalt coating had the positive impact of promoting an increase in the austenite proportion in the weld bead. Ferrite grains covered with allotriomorph grain boundary austenite, Widmanstätten austenite nucleated from allotriomorph grain boundary austenite, and intragranular austenite make up the weld bead microstructure of the NiW and CoW samples. Allotriomorphs, which form at the ferrite grain borders, Widmanstätten side plates, which are developing into the grain from the allotriomorph grain boundaries, and intragranular austenite are the three main forms of austenite that may be found in SDSS weld metals. While idiomorphic austenite forms at lower temperatures, allotriomorphs and Widmanstätten austenite form at higher temperatures (Muthupandi et al., 2003; Mohammed et al., 2017). The effects of electroplating's addition of nickel and cobalt on the

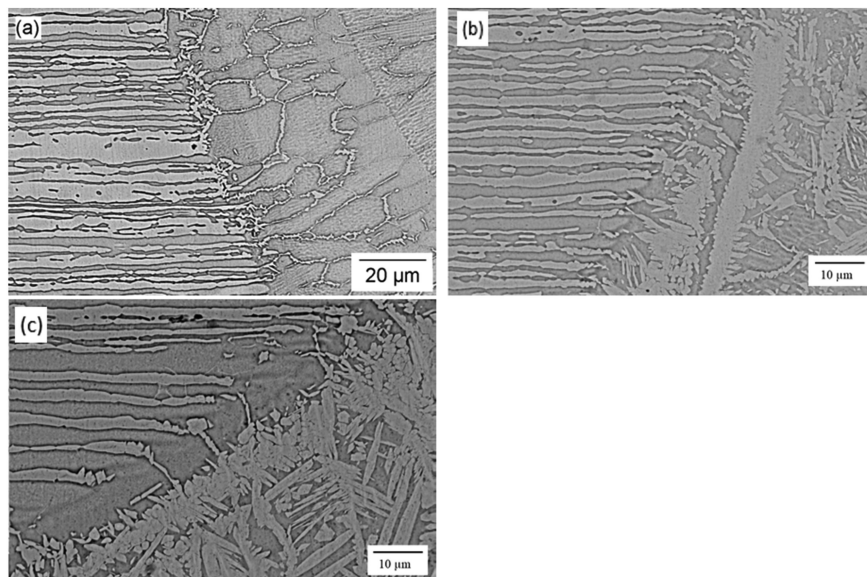


Figure 8.5 SEM images of the transition region between the base material and fusion zone for (a) autogenous welding, (b) NiW, and (c) CoW.

formation of austenite overlapped those of the welding thermal cycle's promotion of an increase in austenite proportions even at high cooling rates.

Figure 8.5 depicts the microstructures under all conditions in the region between the base metal and weld bead. For the NiW and CoW samples, although the amount of austenite in the fusion zone is greater than 50%, in the HAZ the amount of ferrite was higher than that of austenite. Despite not being hot enough to melt the material, the HAZ's temperature changes the phase balance. The HAZ's austenite content is slightly lower than that of the weld center because this area cools more quickly from the temperature range above γ - δ solvus, where ferrite predominates, leaving less time for additional austenite to form during cooling. The unbalanced microstructure will compromise the HAZ's resistance to corrosion.

Table 8.3 illustrates the volume fraction of austenite and for all the settings, and it provides quantitative information about the impact of nickel and cobalt electroplating on the laser welding process. Comparatively, the austenite volume percentage in the NiW condition was around 57%, and the cobalt volume fraction was 48% in the AW condition. The NiW and CoW show good results given that austenite values lower than 25% are not suitable for the majority of industrial applications (Saravanan et al., 2017; Kang et al., 2013). According to some sources, the weld bead must include at least 30% austenite

Table 8.3 Phase balance on weld bead

Phase proportions		
Sample	Austenite%	Ferrite%
AW	7.6 ± 0.5	92.4 ± 0.5
NiW	56.2 ± 0.4	43.5 ± 0.4
CoW	47.9 ± 0.3	52.1 ± 0.3

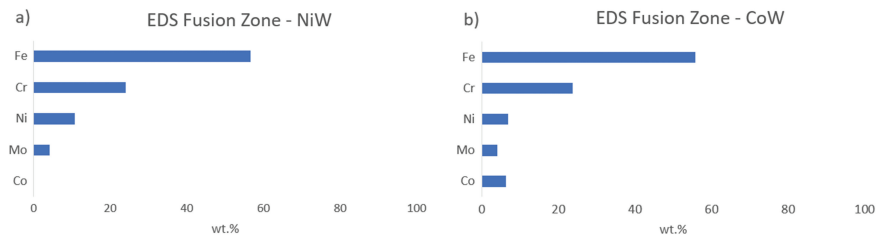


Figure 8.6 Fusion zone composition for (a) NiW and (b) CoW.

in order for the weld joints to be accepted (Ramkumar et al., 2014). Higher austenite (FCC) content produces good corrosion resistance but lower strength, while higher ferrite (BCC) content produces high strength but poor corrosion resistance. Balanced austenite/ferrite proportions are essential for a desirable mix of mechanical strength and corrosion resistance.

When all welding conditions were kept the same for all samples, the NiW and CoW samples demonstrate that Watts bath was effective in resolving the problematic imbalanced microstructure that resulted from SDSS welding. According to S. Saravanan et al. (2017), changing the welding settings alone does not produce acceptable ferrite–austenite proportions for Nd:YAG laser weld joints. The Watts bath encouraged the enrichment of nickel and cobalt in the weld bead, which led to favorable austenite/ferrite proportions. The effect of nickel as an austenite-promoting element is well studied and applied, but the results show that cobalt could also be utilized for the correction of phase balance on the SDSS laser welding.

The nickel–cobalt enrichment, promoted by the Watts bath, can be estimated in Figure 8.6 that shows the composition (wt.%) of the fusion zone obtained by EDS analysis for the samples NiW and CoW (a and b, respectively).

For the NiW sample, the Cr, Fe, and Mo concentrations remained the same for the base metal and the concentration of nickel (fusion zone) increased from approximately 7% to 11%. For the CoW sample, the Cr, Ni, Fe, and Mo concentrations remained the same for the base metal. The cobalt amount in the fusion zone (approximately 6%) was from cobalt coating once the base metal did

Table 8.4 Weld bead hardness

Condition	Fusion zone hardness [HV]
AW	400 ± 2
NiW	320 ± 8
CoW	345 ± 6

not have cobalt in its composition. The increase of nickel and cobalt amount increased the volume fraction of austenite, corroborating the results observed in the micrographs and proving the efficiency of the addition of nickel and cobalt by Watts bath as a method to correct the phase balance on the Nd:YAG pulsed laser welded UNS S32750. Nickel was more efficient in the formation of austenite since a smaller amount was needed, compared to cobalt, to correct the phase balance in the microstructure.

Table 8.4 displays the average microhardness of the weld bead for each condition. The base metal average microhardness was 307 ± 4 HV. The AW sample presents the highest hardness due to the predominantly ferritic microstructure. Higher ferrite content leads to high strength (Tan et al., 2012). As envisaged, a drop in hardness is shown (NiW and CoW) as the amount of austenite in the weld bead increases.

Although for the NiW and CoW conditions, the volume fractions are very close to those of the base material, the microhardness value was higher. The large amount of intragranular austenite and its arrangement along the fusion zone created dislocations thus raising the hardness.

While a high ferrite content boosts mechanical strength at the expense of corrosion resistance, a high austenite content boosts corrosion resistance while lowering mechanical strength. A desired combination of mechanical strength and corrosion resistance requires a balance between the austenite and ferrite fractions.

Conclusion

The following conclusions could be drawn:

- The Watts bath allowed the formation of a nickel and cobalt electrodeposited layer with a constant thickness and well adhered to the base metal surface.
- The use of nickel–cobalt electroplating technique was effective in promoting austenite formation and consequently correcting the resulting microstructure in SDSS laser welding.

- It is feasible to use cobalt as an austenite-promoting element in SDSS welding.
- Nickel was more efficient in the formation of austenite since a smaller amount was needed, compared to cobalt, to correct the phase balance in the microstructure.
- The increase in austenite proportion decreases the microhardness of weld bead. Although the volumetric proportion of austenite in the fusion zone for the NiW and CoW conditions was similar to that of the base material, the hardness was greater because the morphology and distribution of the austenite along the fusion zone had an impact on the hardness of the region.
- The results from the method outlined might be used to broaden the design of the welding settings to include other DSS and laser sources.

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Chapter 9

Laser Welding of Ultra-High Strength Steels for Solid Rocket Motors

H. G. Svoboda^{1,2} and M. A. Duran^{1,2}

¹GTSyCM3—Welding Technology Group, INTECIN, Faculty of Engineering, University of Buenos Aires, Buenos Aires, Argentina

²CONICET, Buenos Aires, Argentina

Abstract

Ultra-high strength materials (1000–1800 MPa) have been used for aerospace and defence applications, in different parts of Solid Rocket Motors (SRM). The case is a critical tubular component fabricated by welding, mainly using AISI 4130 or Maraging steels. Its structural integrity is a relevant aspect, controlled by the quality of the welded joints. In this sense, the optimization of the welding procedure is important to assure the established requirements.

AISI 4130 is a Cr-Mo steel with medium C content, widely used for small SRM cases, which reach between 900 and 1200 MPa of tensile strength, with good toughness in quenching and tempering (Q&T) condition. Its weldability is limited by the relatively high alloy content. Due to the welding thermal cycle produces a microstructural gradient through the welded joint. A post-weld heat treatment (PWHT) is required to adjust the final mechanical properties.

Maraging C250 steel has an extra-low C content (<0.01%) and 18% Ni as a main alloying element, reaching an ultra-high strength (1500–1800 MPa) in the ageing condition, with excellent toughness, dimensional stability and weldability, being a very good option to produce welded SMR cases. After welding, it also has a microstructural gradient, which is mainly homogenized after a PHWT of ageing.

Laser Welding (LW) and Gas Tungsten Arc Welding (GTAW) are the main welding processes used for this application. Also, Electron Beam Welding (EBW) or Friction Stir Welding (FSW) could also be applied.

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Due to many aspects as high energy density, automatization and high welding speed, among others, LW is one of the most recommended processes. Nevertheless, there is still a lack of information related to the effect of the welding/manufacturing procedure on the microstructural evolution, defects and resultant mechanical properties. In the last years Additive Manufacturing (AM) has revolutionized the manufacturing industry, and nowadays Laser Metal Deposition (LMD), which is closely related to LW, is considered a process that could be used for case fabrication.

The objective of this chapter is to analyse the effect of Laser Welding procedure of AISI 4130 and Maraging C250 steel sheets on the microstructural evolution and resultant mechanical properties of the obtained welding joints, used for the fabrication of SRM cases. The implementation of LMD for the case, as well as other structural components of SRM, are also presented and discussed.

Keywords: Laser welding; Solid Rocket Motor (SRM) Case; Maraging steels; AISI 4130 steel; Laser Metal Deposition (LMD); Microstructure; Mechanical properties

Introduction

Solid Rocket Motors (SRM)

In rocket propulsion systems for aerospace applications there are two main types: Liquid Rocket Engines and Solid Rocket Motors (SRM). Rocket propulsion is a class of jet propulsion that produces impulse by ejecting matter (propellant), stocked entirely in the flying vehicle, which does not require oxygen from the atmosphere for the combustion of it. The energy for the thrust is achieved by a chemical reaction. In SRM the propellant located inside the combustion chamber or case is solid and it is called “grain” (Sutton and Biblarz 2017). They are widely used to provide a wide range of thrust from initial spacecraft launch to missile applications (NASA 1970), having the advantage of simplicity, compactness, easy maintenance in the full lifecycle and high thrust for relatively low cost, making them a very competitive propulsion system (Brill et al. 2000, Zou et al. 2021, Wang et al. 2022). High pressure and high temperature gases are generated as a consequence of the combustion reaction inside the case and the convergent-divergent nozzle (Kumar et al. 2016, Singh 2013, Evans 1988). The case acts as a highly loaded pressure vessel, being the most important structural component of the SRM (Evans 1988). In Table 9.1 are listed some of the more relevant loads to which SRM cases are submitted during the flight (Sutton and Biblarz 2017).

Table 9.1 Main loads in SRM case.

Origin of load	Type of load/stress
Internal pressure	Tension biaxial, vibration
Axial thrust	Axial, vibration
Rocket motor nozzle	Axial, bending, shear, thermal
Thrust termination equipment	Biaxial, bending
Thrust vector control actuators	Axial, bending, shear
Aerodynamic control surfaces	Tension, compression, bending, shear, torsion
Staging	Bending, shear
Flight manoeuvring	Axial, bending, shear, torsion
Vehicle mass and wind forces on launch pad	Axial, bending, shear
Dynamic loads from vehicle oscillations	Axial, bending, shear
Start pressure surge	Biaxial tension
Ground handling	Tension, compression, bending, shear, torsion
Ground transportation	Tension, compression, shear, vibration

Also, during the operation of the SRM, the case experiments a continuous increasing of temperature at their walls, due to the heat transfer from different heat sources (propellant combustion, atmosphere, etc.). This material heating could produce material degradation and additional thermal stresses (Sutton and Biblarz 2017).

Figure 9.1 shows a schematic diagram of SRM, whose dimensions and design depend on rocket motor requirement application (Sutton and Biblarz 2017; Singh 2013). The design and fabrication technology of cases have progressed in the last time introducing new materials and manufacturing technologies improving the SRM performance.

Materials and Manufacturing Technologies

Cases are usually made either of metallic alloys or with wound-filament composite materials of polymeric matrix or with combinations of both types of materials (Sutton and Biblarz 2017, Kumar et al. 2016, Shamantha et al. 2000).

Considering metals, the materials typically used are High Strength Steels, Aluminium alloys and Titanium alloys.

For fibre-reinforced polymers the strength-to-density ratio is higher than metallic alloys; nevertheless, they have some important limitations. Metallic cases have several advantages such as supporting higher temperatures (700–1000 °C). Also, they usually show ductile behaviour before failure and are less

sensitive to rough handling. Indeed, they do not deteriorate with weather or time exposure and are more tolerant to concentrated loads. They usually could carry more propellant (Sutton and Biblarz 2017).

High-strength steels have been most commonly used for case fabrication, but other materials like aluminium, titanium, and nickel alloys have also been used. Table 9.2 gives a comparison of SRM case material properties (Sutton and Biblarz 2017).

Among high and ultra-high strength steels, Cr-Mo medium carbon alloyed steel (AISI 4130) and lastly maraging steels (MA) are the most widely used steels for SRM case fabrication.

AISI 4130 is a 0.30 wt% steel, with low contents of Cr and Mo as alloying elements which is used in a quenching and tempering (Q&T) condition, reaching tensile strength (σ_{UTS}) levels of 1000–1200 MPa. Maraging steels (MA) are an ultra-high strength family of materials which can reach up to about 1900 MPa in combination with good fracture toughness. These alloys in an annealed condition consist of a relatively low-hardness martensite (approx. 300 HV), reaching the final strength values from ageing at relatively low temperatures (450–500 °C) (Passarelo Moura da Fonseca et al. 2021; Tewari et al. 2000; Floreen 1968).

For metal case manufacturing the joint technology commonly accepted is welding (Figure 9.1) which must meet aerospace standards, ensuring the quality of the structural component (Sundarsan et al. 2010; AWS D17.1 2010; Tsay et al. 1992). Welding should introduce as low heat input as possible, generating minimum distortions and high-quality welds. Laser Welding (LW), Electron Beam Welding (EBW) and Plasma Arc Welding (PAW) are processes with high power densities, which produce a narrow weld metal (WM) and heat-affected zone (HAZ) (Tsay et al 1992; Joshi et al. 2016; Murthy et al. 2019; Norrish 2006). Generally, it does not need joint preparation nor filler material. Friction Stir Welding is a solid-state welding process which could be also used

Table 9.2 Physical properties of metallic materials typically used for SRM, at 20 °C.

Material	Tensile strength [MPa]	Modulus of elasticity [GPa]	Density [g/cm ³]	Strength to density ratio [1000]
Titanium alloy	1240	110	4.6	270
Alloyed steel*	1000–2000	207	7.84	128–255
Aluminium alloy*	460	72	2.79	165

*In heat-treated condition

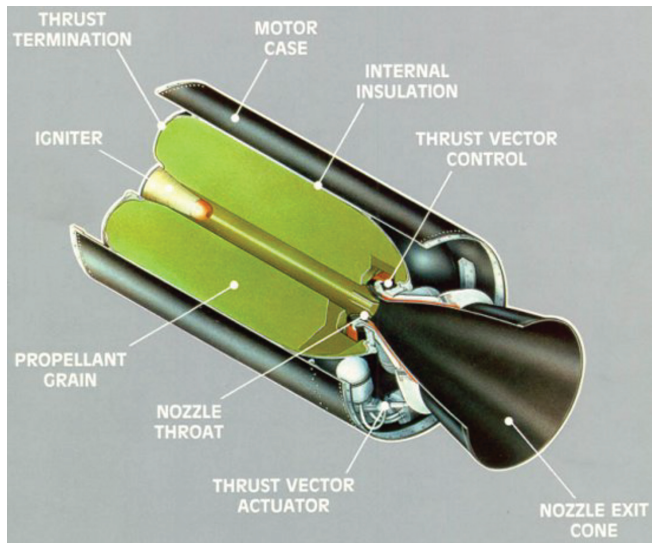


Figure 9.1 Schematic of SRM (Kumar et al. 2016).

for these applications (Li et al. 2021). Post-weld heat treatment (PWHT) is usually needed to adjust the final properties of welded joints (Tsay et al. 1992).

In the last years Additive Manufacturing (AM) has become a revolutionizing technology enabling several improvements in many industries, with particularly impact in aerospace applications (Blakey-Milner et al. 2021). Related with 3D printing of metallic components, there are three main processes with high maturity: Selective Laser Melting (SLM), Laser Metal Deposition (LMD) and Wire Arc Additive Manufacturing (WAAM). All of them are strongly related with welding, producing metallic parts from powders or wires, melted and deposited on a predetermined path, layer by layer (Freeman 2021). Particularly, LMD is a Direct Energy Deposition (DED) process (Figure 9.2), related to Laser Welding or Laser Cladding that allows to produce large metallic parts (Svetlizky et al. 2021). Particularly, LMD can be used to build up tubular cases and nozzles for SRM, among other components (Blakey-Milner et al. 2021).

Motivations and Objectives

The aim of this chapter is to review the state of the art of laser welding of ultra-high strength steels used in the fabrication of SRM cases. Materials, welding parameters, defects and microstructural evolution, post-weld heat treatments and mechanical properties of the welded joints are some of the

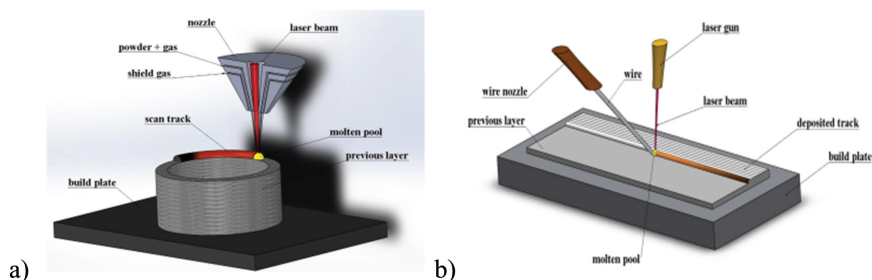


Figure 9.2 Schematic of LMD processes: a) with powder, b) with wire (Moeinfar et al. 2022).

aspects analysed. Laser Metal Deposition (LMD), which is closely related to Laser Welding, as an actual and future technology for SRM case fabrication is presented and discussed. Remarks and future outlook are also considered.

Ultra-High Strength Steels for SRM case

Ultra-high strength steels have been the metallic material mostly employed in SRM case fabrication. These materials allow the use of thinner thickness without significantly changing the service/design requirements. Cr-Mo steel (AISI 4130) has been historically widely used steel in aircraft and small SRM case construction because of its low cost and good combination of high strength and toughness in quenching and tempering (Q&T) condition (Tsay et al. 1992; Rajan et al. 2002; Joshi et al. 2016). On the other hand, MA steel has received considerable attention for exhibiting a higher combination of properties such as ultra-high strength with excellent fracture toughness and weldability (Sudarsan et al. 2010; Sha and Guo 2009). Chemical composition, microstructure, heat treatments and mechanical properties of these materials are presented in this section.

Cr-Mo Steel (AISI 4130)

This Cr-Mo steel is recognized for its good formability in annealed condition (Rajan et al. 2002) as well as for its acceptable ductility in Q&T (Lim et al. 2022) condition which promotes its use in the manufacturing of pressure vessels. Emamian et al. (2010) have pointed out that the higher specific strength achieved enables to reduce the weight of designs by using thinner thicknesses and ensure the structural strength component.

Table 9.3 shows the chemical composition of AISI 4130 steel (ASM Handbook 2005).

It is a medium carbon steel alloyed with Cr (1 wt%) and Mo (0.20 wt%). In annealed condition, the material has a ferritic-pearlitic/bainitic microstructure typically slightly banded produced during hot-rolling (Figure 9.3a).

In Figure 9.3b a detail of the carbides heterogeneously distributed in the ferritic matrix and grain boundaries can be seen.

Emamian et al. (2010) stated that according to the chemical composition of AISI 4130, heat treatment conditions and CCT diagram, it is possible to obtain a ferritic-pearlitic, ferritic-bainite and full martensite structure. The alloy content was established to provide a moderate quenchability, to obtain a full martensitic structure after an oil quenching (Lee et al. 2012), after which a tempering is necessary to adjust the final mechanical properties. The tempering temperature usually varies between 450 at 650 °C approximately (ASM Handbook 1991).

After quenching, lath martensitic structure is formed (Figure 9.4) which is hard and brittle. Carbon supersaturation of the martensite is the driving force for carbide precipitation during tempering, with the consequent depletion in C of the matrix.

Precipitation takes place preferably in the residual dislocations of the martensite or at grain boundaries. Finally, the microstructure after tempering

Table 9.3 Chemical composition of AISI 4130 steel (wt%).

Alloy	C	Cr	Mo	Mn	Si	S _{max}	P _{max}	Fe
4130	0.28–0.33	0.8–1.1	0.15–0.25	0.4–0.6	0.15–0.35	0.04	0.03	Bal

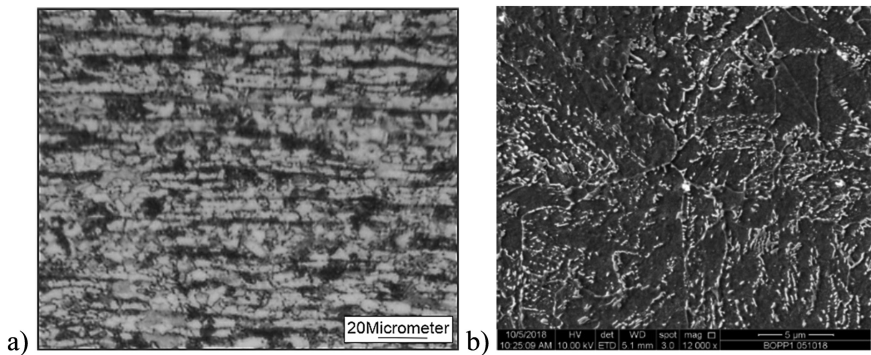


Figure 9.3 Microstructure of AISI 4130 in annealed condition. a) Light Microscopy (LM) and b) Scanning Electron Microscopy (SEM).

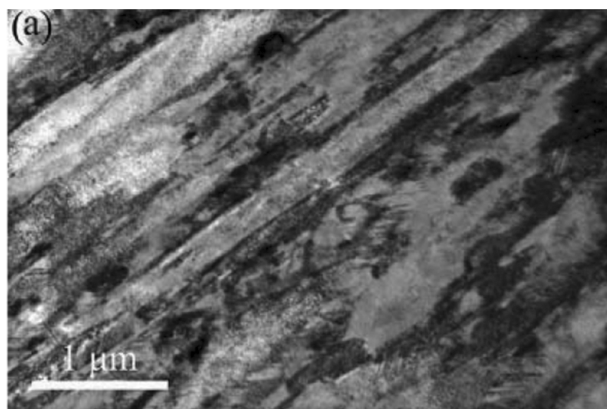


Figure 9.4 TEM image of lath martensite obtained from quenching in oil a Cr-Mo-Nb-B steel with 0.25 wt% of C (Zheng et al. 2019).

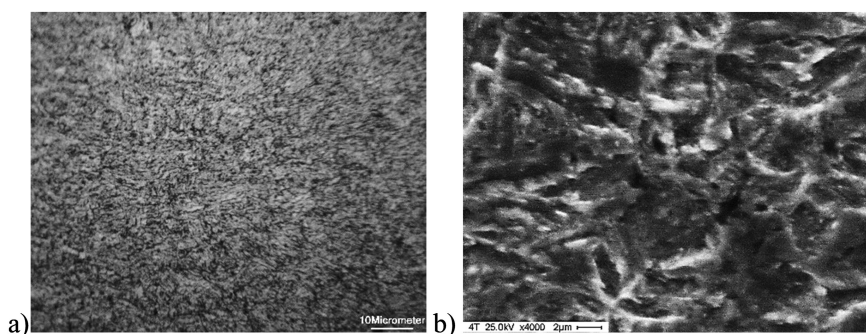


Figure 9.5 Microstructure of AISI 4130 in Q&T (520 °C-1h). a) LM, b) SEM.

was composed of finer alloyed carbide dispersion such as Cr_{23}C_6 and Mo_6C in minor proportion, in a tempered martensite (ferritic) matrix, with an average size between 0.5 and 1 μm (Park et al. 2010). Figure 9.5 shows the microstructure obtained after Q&T.

The final hardness, yield and ultimate tensile strength decrease with increase in temperature due to carbide coarsening and recovery mechanisms. Nevertheless, the replacement of coarse cementite by a finer alloyed carbide could produce a slight increase in hardness near 450 °C (Porter et al. 2022). Table 9.4 shows the range of mechanical properties that can be obtained depending on the tempering temperatures (ASM Handbook 2005; Duran et al. 2021; Oh et al. 2022).

AISI 4130 used for SRM case manufacture usually has a tensile strength greater than 1000 MPa (Sutton and Biblarz 2017). Duran et al. (2021) obtained

Table 9.4 Mechanical properties of AISI 4130 steel after Q&T.

T _{tempering} [°C]	UTS [MPa]	YS [MPa]	Elong. (%)	HV
400 – 550	1400 – 900	1300 – 800	5 – 13	450 – 300

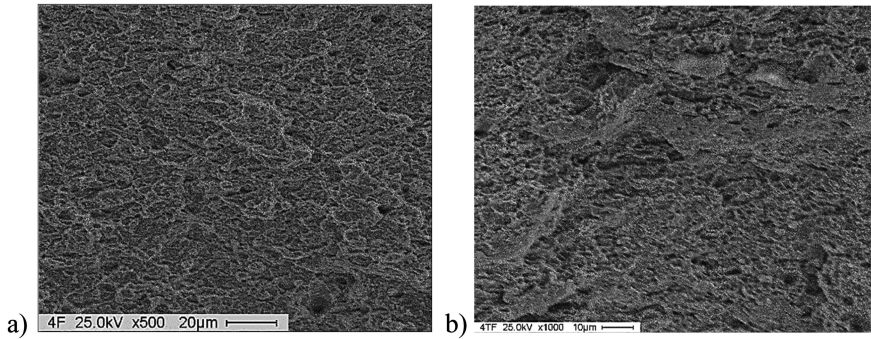


Figure 9.6 Ductile fracture surface of AISI 4130 in Q&T condition.

σ_{UTS} =1000 MPa in a 2 mm thick sheet of AISI 4130 after austenitized at 840 °C per 20 min, oil quenched and then tempered at 520 °C per 1 h, and 1200 MPa when the tempering temperature was 440 °C, with hardness in the range of 320–410 HV. The obtained elongations were between 6 and 9%. Oh et al. (2022) have proposed a toughness estimation method (K_f) obtained from tensile test results, resulting in K_f =212 MJ/m³. Antonelli Martiniano et al. (2021) have reported for AISI 4130 after austenitized above 850 °C per 1 h, oil quenched and tempered at 553 °C per 1 h for a tensile strength superior to 1200 MPa. Furthermore, they stated the material showed a ductile fracture surface, characterized by coalescence of microvoids. Figure 9.6 shows a fracture surface morphology for an AISI 4130 in Q&T condition.

Carney et al. (2013) have mentioned that this mechanics of fracture is representative of this material type and depends on the matrix nature and the distribution of the second phases, such as carbide and inclusion. The dimples are formed by nucleation, growth and coalescence of microvoids and their size could be related to fracture toughness.

Maraging Steel

Maraging steels are a special class of ultra-high strength steel which are hardened by reactions of precipitation that do not involve carbon, produced during ageing heat treatment (ASM Handbook 2005). These alloys contain extra low

carbon content with high levels of Ni, Co and Mo with a small but significant amount of Ti (Floreen 1968). There are several grades of commercial MA steels the most widely used and most commonly available grades are those of the C Series, which contains Co, with strengths in the range of 1300 to 2400 MPa (Christopher et al. 2004; Shan and Guo 2009). Table 9.5 shows the different chemical compositions of the C grade maraging steels.

C350 grade is an ultra-high strength made in limited quantities for special applications. Its strength is mainly associated with the higher Ti content.

Cobalt is an expensive alloying element. This aspect maintains series C of MA steels rather expensive. In this sense, in the last years the development of a cobalt-free maraging steel of strength levels up to 2000 MPa was a global objective (He et al. 2002). This variant is identified as series T due to the use of Ti. Table 9.6 shows the chemical composition of T grades of MA steels (ASM Handbook 2005; Sha and Guo 2009).

These alloy grades were developed to replace the Co for Ti and reduce the cost of the material having similar mechanical properties (Li et al. 2016). Up to now, the higher Ti content can only be carried out on T200 and T250 alloys (Pan et al. 2019).

All the maraging grades can be considered in the same way (Sha and Guo 2009). Due to the high Ni content of the alloy, when it is cooling from the austenitic field, it does not transform into the equilibrium phases (austenite + ferrite). Instead, it does into martensite during cooling to room temperature (annealed condition) (Floreen 1968). Kumar et al. (2017) mentioned that the

Table 9.5 Chemical composition of C grades of MA steels in W% (ASM, Handbook 2005).

Grade	C	Ni	Co	Mo	Ti	Al	Fe
C200	<0.03	18	8.5	3.2	0.2	0.1	Bal
C250	<0.03	18	8.5	5.2	0.5	0.1	Bal
C300	<0.03	18	9	5	0.7	0.1	Bal
C350	<0.03	18	12.5	4.2	1.6	0.1	Bal

Table 9.6 Chemical composition of T grades of MA steels in W%.

Grade	C	Ni	Co	Mo	Ti	Al	Fe
T200	<0.03	18.5	-	3	0.7	0.1	Bal
T250	<0.03	18.5	-	3	1.4	0.1	Bal
T300	<0.03	18.5	-	4	1.85	0.1	Bal

martensite formed has a BCC (body-centred cubic) structure with high dislocation density and is non-twinned and is presented in the form of packets or laths (Passaro Moura da Fonseca et al. 2022) (Figure 9.7).

The material could be aged at different low temperature heat treatments from 400 °C to 600 °C per 3–5 hs (Passaro Melo da Fonseca et al. 2022). The hold time and temperature depend on the MA grade (Sha and Guo 2009). For C250 the best combination of strength and toughness is achieved at 480 °C per 3 hs, being the standard condition (Samantha et al. 2000).

During ageing treatment, it takes place the precipitation of nano-sized Ni_3Mo , Ni_3Ti and Fe_2Mo intermetallic compounds, distributed in the dislocated martensite matrix. They are the responsible for the hardening of MA steels (Mshka et al. 2015; Kumar and Ghosh 2017). The morphology, size and volume fraction of the precipitates are affected by the chemical composition of the alloy and the heat treatment parameters (Li et al. 2022; Tewari et al. 2000). In the case of the cobalt-content grades, the main contribution of this element is to decrease the solubility of Mo, promoting the formation of Ni_3Mo precipitates during age treatment (ASM, Handbook 2005; Sha and Guo 2009). Also, martensite reversion reaction to austenite and ferrite could occur during heat treatment (Hamilton et al. 2015). Nevertheless, the precipitation reactions take place much faster (Sha and Guo 2009).

The small size of the precipitates (<30 nm) difficult the use of conventional characterization methods such as BFTEM (Bright Field Transmission Electron Microscopy), EDS (Energy Dispersive Spectrometry) or low magnification SEM (Scanning Electron Microscopy). Moshka et al. (2015) have

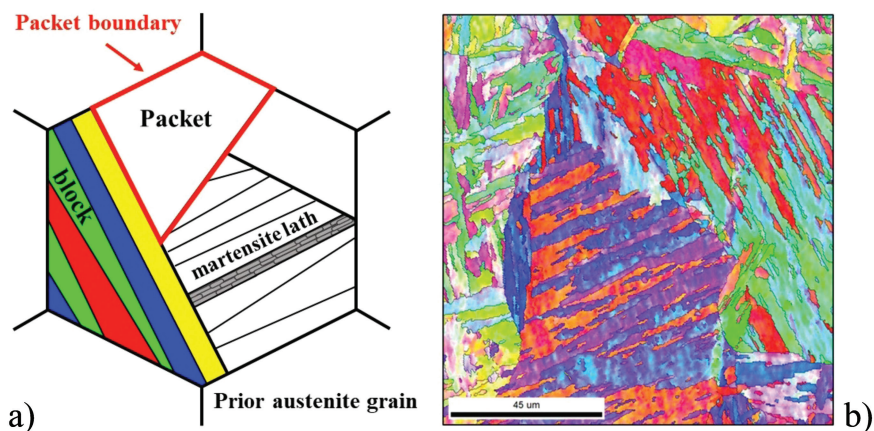


Figure 9.7 Martensite in MA steel: a) Schematic, b) EBSD map of C350 (Passaro Melo da Fonseca et al. 2022).

characterized the nano-size precipitate by DF-STEM (Dark Field Scanning Transmission Electron Microscopy), STEM-EDS mapping and SAED (Selected Area Electron Diffraction) patterns. They have reported that the Ni_3Mo precipitate has a needle-like morphology, while the Ni_3Ti has an ellipsoidal one. For higher aging times, Ni_3Mo is dissolved and Fe_2Mo is formed, producing a Ni enrichment of the matrix, which allows the formation of reverted austenite. For a C300 steel, Dehgahi et al. (2020) and Mei et al. (2022) have reported a greater fraction of spherical Ni_3Ti as a consequence of the increased Ti alloy content and also an increase in reverted austenite fraction at high aging temperatures. Figure 9.8 shows a maraging steel microstructure in annealed and aged conditions.

The martensite reverse reaction is one of the most studied issues for all maraging steel grades, although it usually does not occur at industrial conditions of heat treatment. When the material is reheated at temperatures below the A_{C1} or during prolonged times (>40 hs), the martensite tends to decompose to the equilibrium phases. Abreu et al. (2015) stated that the austenite formation is a result of Ni-enrichment due to precipitate coarsening during prolonged times at ageing temperatures. The austenite enriched in Ni remains stable on cooling to room temperature. It has been reported that at temperatures where the tendency to reversion is high, it may precede the precipitation reaction as a consequence of partitioning of alloying elements (Shamantha et al. 2000; Sha and Guo 2009).

Gomes de Carvalho et al. (2012) have studied the kinetics of martensite-to-austenite transformation by dilatometry analysis. They proposed that the transformation of martensite to austenite can be controlled by a diffusion mechanism at temperatures near A_{C1} and lower heating rates (martensite reversion reaction) and by shear at temperatures near A_{C3} and high heating rates (Kapoor et al. 2003).

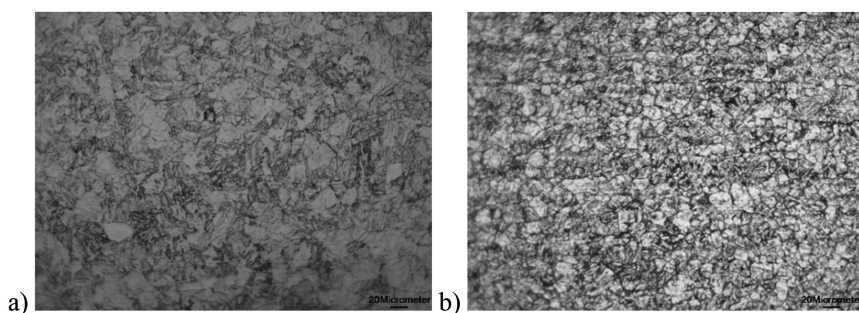


Figure 9.8 Microstructure of MA: a) Annealed condition, b) Aged condition (480 °C-3hs).

There are two types of reverted austenite, i.e., lamellar, nucleated in the lath martensite boundary, and acicular, nucleated inside the martensite lath (Abreu et al. 2015). According to the literature, many authors have used Electron Back Scatter Diffraction (EBSD), X-Ray Diffraction (XRD), Ultra-High-Resolution SEM, EDS, SAED and BFTEM to characterize it (Sha and Guo 2009; Dehgahi et al., 2021; Li et al. 2022).

After ageing treatment, the hardness increases considerably from 300–320 HV to 520–560 HV (Tariq et al., 2009; Sha and Guo 2009; Floreen 1968).

The ultra-high strength is achieved by the hardening nano-sized precipitates (Ni_3Mo , Ni_3Ti and Fe_2Mo) formed in the dislocated martensite matrix (Moshka et al., 2015). Pan et al. (2019) and Li et al. (2022) have mentioned that although for maraging T type it is possible to achieve a similar strength that of C types, due to the high Ti content the ductility of the material may be limited (Li et al. 2019). Li et al. (2022) have shown that the beneficial contribution of nanometer-scaled lamellar reverted austenite in the improvement of ductility in T250 steel.

Duran et al. (2021) have studied the mechanical properties of a C250 grade for two ageing temperatures: 480 °C and 520 °C, both with 3 hs of permanence, obtaining $\sigma_{\text{UTS}}=1640$ MPa and $\sigma_{\text{UTS}}=1770$ MPa, respectively. Similarly, Shamantha et al. (2000) have reported for the same MA grade and soaking time, $\sigma_{\text{UTS}}=1500$ MPa for 440 °C, and for 480 °C and 520 °C values higher than 1700 MPa. The elongation values are in the range of 8–10% (Tariq et al. 2009; Duran et al. 2021; Shamantha et al. 2000).

Sha and Guo (2009) have pointed out that the failure surface of tensile specimens for the maraging steels exhibits dimples, typical of the ductile fracture mechanism. In the early aged and annealed condition, the dimples are big and deep. Figure 9.9 shows a fracture surface morphology of a tensile tested sample of C250, aged at 480 °C – 3 hs.

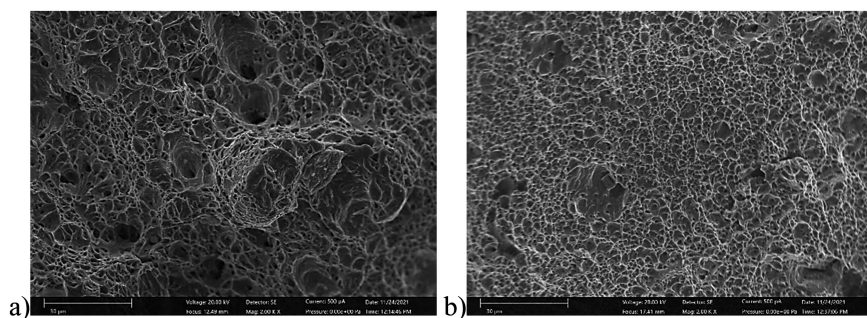


Figure 9.9 Ductile fracture surface of tensile tested sample of C250 steel.

a) Annealed condition and b) aged condition (480 °C-3hs).

When ageing time and temperature increase, the dimples become uniform and shallow as a consequence of the increase in the precipitate fraction. The same authors have reported that when ductility decreases, it reduces the fracture surface energy, resulting in lower fracture toughness. For a T250 in annealing condition it was reported 98 MPa m^{1/2} and 73 MPa m^{1/2} after ageing at 480 °C-3 hs. Wang et al. (2020) have studied in C350 the effect of ageing temperature on fracture toughness, reporting a decrease of it when the resistance increases. Also proposed is the presence of micro-toughness cleavage plane which tends to disappear when the material is aged. For prolonged aged state, the fracture toughness can be improved due to a slight formation of lamellar austenite and the increase of the precipitate spacing. They reported 76 MPa m^{1/2} at 500 °C-5 hs and 166 MPa m^{1/2} at 600 °C-3hs.

Shamantha et al. (2000) studied the fracture toughness of a C250 in different conditions. For ageing at 440 °C for 3 hs has reached high K_Q values (132 MPa m^{1/2}) but with relatively low strength (<1500 MPa). The material achieved the established tensile strength at 480 °C after 3 hs with a slight decrease in K_Q values (117 MPa m^{1/2}). They have pointed out that C250 grade has a fracture toughness-strength ratio greater than other maraging steel grades. This is one of the reasons for the wide use of this grade in SRM case manufacturing.

Figure 9.10 presents a resume of results of fracture toughness vs. UTS strength for different MA steel grades (Passaro Melo da Fonseca et al. 2022). The decrease of it could be seen with the strength's increase.

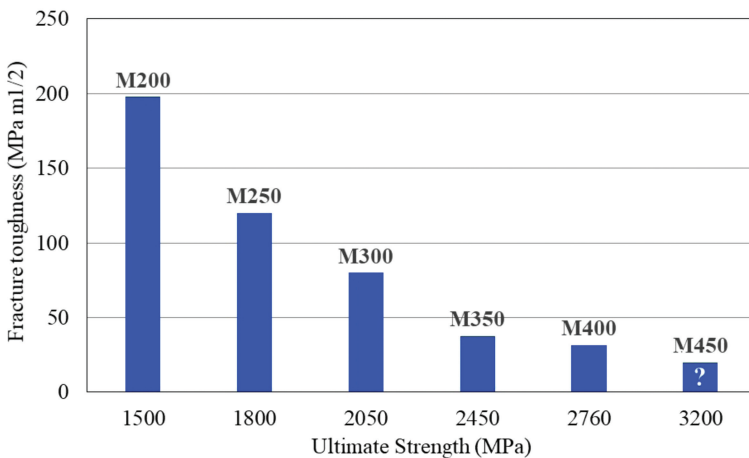


Figure 9.10 Fracture toughness for different UTS strength levels of MA steels in peak hardening condition (Passarelo Moura da Fonseca et al. 2022).

Laser Welding of Ultra-High Strength Steels for SRM case

Laser beam welding (LW) has been widely used in many applications in different industries. This technology is capable of producing high-quality welds, minimizing not only the heat input but also the distortion of weldments (Wu 2003). Also, lower requirements for joint preparation are needed (butt joints), and generally filler materials are not used. Consequently, it is very useful for the aerospace industry, particularly for SRM case manufacturing.

In this application, circumferential sectors of relatively thin sheets (2–3 mm thickness) are conformed and longitudinally welded. Then, several sectors are circumferentially welded to construct the case (Figure 9.1). In order to study the influence of different LW process parameters on the defect generation, microstructural evolution and final mechanical properties, bead on plate (Li et al. 2017) or butt welds of plain sheet samples (Joshi et al. 2016) are usually employed.

Considering the laser type, CO₂ lasers have been widely used for welding in industrial applications. More recent fibre lasers have replaced the traditional or CO₂ ones due to their flexibility in delivering the laser (Nayak et al. 2015). Yb fibre sources emit laser radiation of approximately 1 µm in wavelength, which can be transmitted down by optical fibres of several metres long to a head that focuses the beam for welding by means of a series of optics. Therefore, the laser source could be far away from the welding zone, and the flexible optical fibre allows to mount the process head to an industrial robot or CNC machine to weld components with complex geometries. On the other hand, laser radiation emitted by CO₂ laser sources is approx. of 10 microns in wavelength, which cannot be focused down by optical fibers, and must be directly applied on the workpiece using a series of optical components, limiting the component geometries that can be welded with this type of laser (Blackburn 2021). Another limitation of CO₂ laser for welding is the formation of an Ar plasma, which reduces weld penetration depth, which can be easily formed in Ar shielding gas. High proportion of He in the shielding gas (>50%) is needed to avoid the plasma formation and produce a deep penetration weld (Katayama 2013). Figure 9.11 shows schemes of both CO₂ and Yb fibre laser equipment.

High power, high electrical efficiency and high beam quality are achieved by fibre lasers. They could be operated in continuous wave (CW) or pulsed wave (PW) mode. PW-produced spots that depend on the pulse characteristics could form a weld bead. Deep penetration welds could be easily produced increasing the pulse width near the focal point of a focusing lens. The maximum depths of laser spot welds should normally be around 1.5 mm or 3 mm under special conditions. Porosity could be easily produced especially when the irradiation time is short and deep welds are made with a rectangular pulse

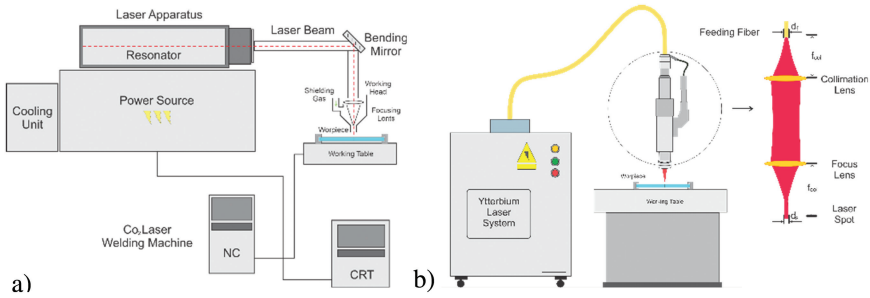


Figure 9.11 Schematic of a) CO₂ laser, b) Yb fibre laser equipment.

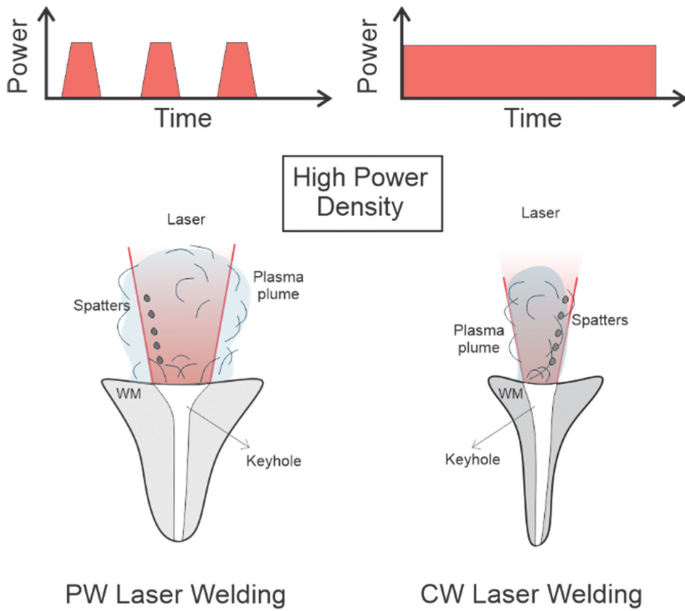


Figure 9.12 Schematic of PW and CW laser welding.

shape. Pulse shaping of slowly rising and falling power is used to reduce spattering and porosity, respectively (Katayama 2013). Figure 9.12 shows schematically typical phenomena in PW and CW laser welding for keyhole-type penetration (Katayama 2013).

Depending on the power density and laser irradiation time, a melt pool is formed, and the morphology could be keyhole or conduction type. When a laser beam strikes on a metallic sheet the temperature of its surface increases laser energy. The laser absorption rises considerably with an increase in the temperature above the melting temperature, and the temperature of the

laser-irradiated area increases higher to the boiling temperature. A cavity (keyhole) is then produced due to evaporation (Katayama 2013).

The main LW variables are laser power (P), welding speed (v), focal spot diameter (d), defocusing (Df) and shielding gas (type, flow rate, side), particularly for these materials. If the laser is pulsed, frequency, pulse duration and waveform are also relevant (Katayama 2013). In Table 9.7 details of different laser welding parameters used for MA and 4130 steels are shown.

For continuous lasers the power varied between 1.8 and 3.7 kW depending on the sheet thickness and other welding parameters (mainly welding speed and spot diameter). Generally, for lower sheet thicknesses defocusing is not used.

Typically, Ar (Li et al. 2016), He (Wu 2003) or Ar+He mixes (An et al. 2012) are used for both sides to protect the material from oxidation reactions. Flow rates varied between 20 and 30 L/min for the front side and similar or lower for the back side (5–20 L/min). Li et al. (2018) used a Cu backing to increase the cooling rate of the welded joint.

There was found scarce literature related to systematic studies on the effect of laser welding parameters on the defect level, microstructure and mechanical properties of welded joints. In the cited publications the welding parameters used were selected based on preliminary tests. Consequently, in those cases the defect level was low. Nevertheless, defects like porosity, spattering, undercutting, lack of fusion, and oxidation, among others, could be observed when the parameters are not correctly selected. For thin sheets the tendency to porosity formation is lower, but undercutting or lack of fusion could be the

Table 9.7 LW parameters used for MA and AISI 4130 steels.

Material	Thickness (mm)	Laser type	Power (W)	Welding speed (m/min)	Spot diameter (mm)	Defocusing (mm)	Ref
T250	2	Fibre	2000	3	0.13	0	Li 2016
SS-MA	3	CO ₂	3000	2	0.6	0	An 2012
T200	3.4	ND	2500	0.75	ND	0.5	Wu 2003
C300	3.3	Fibre	1800	0.18	0.1	0	Sakai 2015
C250	2	Fibre-P	2200/	0.1	0.6	0	Duran 2022
4130	1	Fibre	750	3.6	0.1	0	Souza 2015
4130/C250	2	Fibre	3700	2	ND	ND	Joshi 2016
4130	3.7	CO ₂	2600	0.75	ND	0.5	Tsay 1992
4130	2	Fibre-P	2000/400	0.06	ND	0	Duran 2022
4130	2	CO ₂	1700–2200	1.2–2	ND	-0.2–0.2	Reddy 2020

*ND: Not defined

typical defects. It has been reported that porosity could be easily formed at low welding speeds. Humping or spattering leading to underfilling could occur at high speeds, depending on the smaller or wider beam diameter, respectively (Katayama 2013).

Figure 9.13 shows a butt weld sample of C250 in a sheet of 2 mm thick, obtained using LW (Duran et al. 2022). The good surface quality as well as the small size of the weld bead can be seen.

Reddy et al. (2020) studied LW of AISI 4130 steel in 2 mm thick, analysing the effect of several welding process variables like power, welding speed, beam angle, focal point position (defocusing) and focal length (spot diameter) on mechanical properties of the welded joints, based on Taguchi methodology. They concluded power is the more relevant LW parameter affecting mechanical properties of welded joints. Other parameters also influence the mechanical properties but with a minor effect.

Duran et al. (2022) have studied the effect of welding speed ($v = 6, 10$ and 12 cm/min) and defocusing (Df : 0 or 1 mm under surface) in pulsed-laser welds of a 2 mm thick sheet of AISI 4130, maintaining the other parameters. Figure 9.14 shows macrographs of cross-sections of the welds.

The increase in welding speed produced a reduction in the size of the joint (WM and HAZ), due to the decrease of the heat input, that could be related to the laser energy density $E_d = P \cdot d / v$ parameter. For higher welding speed (12 cm/min) the penetration was incomplete. Also, the variation of defocusing from surface (0 mm) to 1 mm under surface produce a slightly more uniform through-thickness fusion, due to the localization of the focus in the mid-thickness of the sheet.

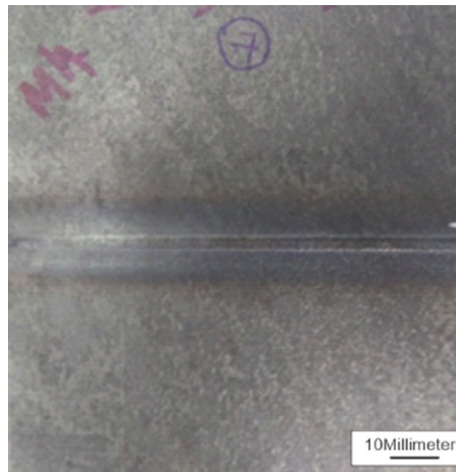


Figure 9.13 MA C250 pulsed laser welded sample for SRM metal case.

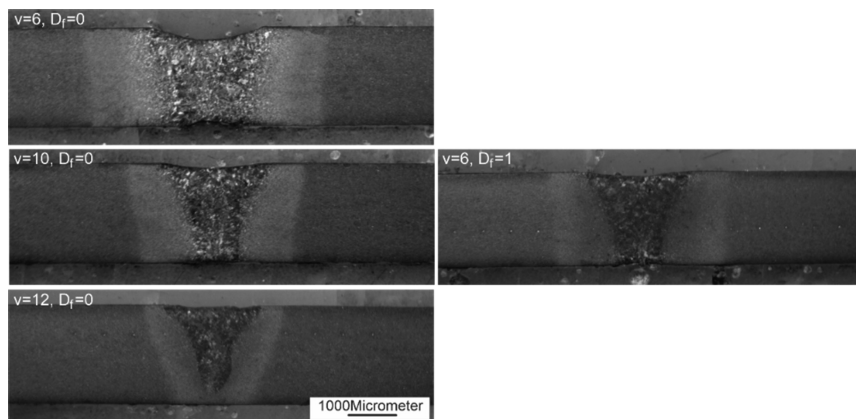


Figure 9.14 Macrographs of pulsed laser welds of AISI 4130 with different welding speed (v) and defocusing (Df) (Duran et al. 2022).

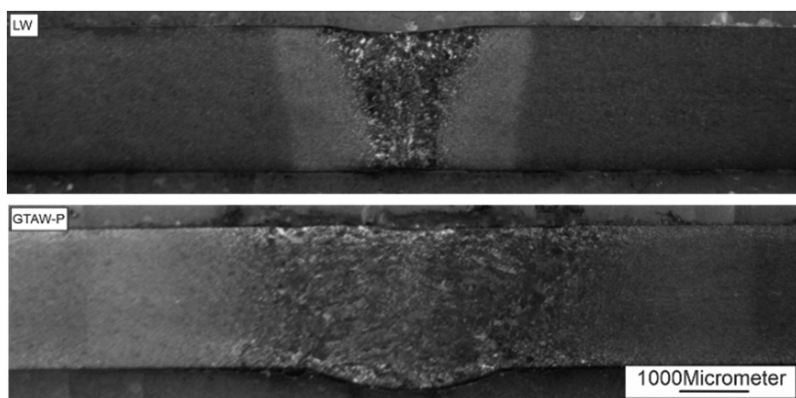


Figure 9.15 Cross-section macrographs of AISI 4130 steel welds. a) LW and b) GTAW-Pulsed.

Several works analyse LW comparatively with other welding processes like GTAW or PAW, both for 4130 and MA steels. GTAW process is widely employed for SRM case fabrication due to the good quality of the obtained welds and its low overall cost; however, this process might produce a bigger HAZ (Souza Neto et al. 2015; Sakai et al. 2015; Duran et al. 2022), due to its intrinsic low power density. Figure 9.15 shows the cross-section macrograph images of both GTAW-Pulsed and LW welds of AISI 4130 steel (Duran et al. 2022).

For some applications in SRM manufacturing, it could be needed to weld dissimilar materials. Particularly AISI 4130 and MA steel are the most used

materials for the case and for internal components also. Joshi (2016) studied dissimilar welding between 4130 and MA steels using a fibre laser, and it was also compared with GTAW welds. It obtained good quality welds, with a high weld joint efficiency for LW joints. The HAZ of LW were smaller than those obtained with GTAW.

Microstructural Evolution in Laser Welded Joints

Regardless of the process, a fusion welded joint is composed of three characteristic zones, i.e., Weld Metal (WM), Heat Affected Zone (HAZ) and unaffected Base Material (BM) (Figure 9.15). During welding, a relatively small part of base material achieves melting temperatures, forming the liquid pool which will form the WM when it cools down. For zones where the temperature reached was below the fusion temperature (HAZ), the material has only undergone solid-state microstructural changes. Finally, the BM has not suffered microstructural changes during the welding cycle. The size of HAZ is influenced mainly by the heat input of the welding process together with the thermal diffusivity of the material.

Several studies have been reported related to the microstructural evolution associated to the weld thermal cycle in both AISI 4130 and MA steels (Joshi et al. 2016; Gupta et al. 2020; Li et al. 2010). Nevertheless, it is still a topic that was not covered extensively. It is an objective of this section to contribute to clarify this subject.

Cr-Mo Steel (AISI 4130)

As Welded (AW) Condition

During welding an AISI 4130 for an aerospace structure, the thermal cycle generated will strongly affect the microstructure obtained in the different zones, defining the resultant mechanical properties of the welded joint (Joshi et al. 2016; Nascimento et al. 2010).

The final condition of the SRM case fabricated with this material is Q&T, but welding could be done both in annealed and Q&T conditions. From a metallurgical point of view the most convenient is to weld it in annealed condition and after welding apply the Q&T heat treatment. In this case, the final microstructure of tempered martensite is quite uniform and mostly independent of welding conditions, achieving mechanical properties similar to those of the base material in Q&T condition (Duran et al. 2021). Nevertheless, it is not

always possible to do, depending on part dimensions and distortions produced during heat treatment. The other possibility is to weld it in Q&T condition, considering the need for a Post-Weld Heat Treatment (PWHT) at relatively low temperature to temper the fresh martensite formed during welding. Both analyses are relevant since the manufacturing of an aerospace component has many steps; hence, the AISI 4130 must probably be welded in these two conditions (Sutton and Biblarz 2017).

In the welded joint there is established a peak temperature gradient from the weld centreline to the BM. The microstructural evolution in the material zone that reaches peak temperatures above A_{C3} will be independent of the initial material condition. The WM presents a solidification structure with columnar grains which transform to martensite. Also, microsegregation could be produced in the last liquid to solidify region. In the high temperature HAZ, near to the fusion line, a coarse austenitic grain also transforms entirely to martensite on cooling room temperature. This region is called Coarse Grain Heat Affected Zone (CGHAZ). In both cases, this transformation results from the relatively high quenchability of the material, the chemical composition and the high cooling rates that take place in LW (Joshi et al. 2016). The HAZ width is quite narrow on LW, depending on its dimensions on the welding parameters used, mainly associated with heat input. Into the HAZ, in the medium temperature region, above A_{C3} temperature, the temperature is not enough to cause grain growth, producing a full recrystallized refined grain. This region is called Fine Grain Heat Affected Zone (FGHAZ). Li et al. (2011) have reported that in the FGHAZ the cooling rates are not sufficiently fast to transform to a fully martensitic structure and other phases such as bainite could form during cooling to room temperature. The contiguous zone to FGHAZ is the Inter-Critical Heat Affected Zone (ICHAZ), where the material is reheated between A_{C3} and A_{C1} temperatures. If the starting material is welded in Q&T condition, the microstructure at ICHAZ is formed by over-tempered martensite, fresh martensite, bainite and ferrite. For annealed conditions, the microstructure in the ICHAZ is composed of fresh martensite, bainite and ferrite. The ferritic phases could be formed when the material is heated just above the A_{C1} or be the original ferrite that does not transform to austenite during the thermal cycle, remaining in the microstructure, independently of the cooling rate (Souza Neto et al. 2011; Li et al. 2011). Finally, there is the Sub-Critical Heat Affected Zone (SCHAZ), where the material has been reheated below A_{C1} temperature. This zone is only relevant when the material is welded in Q&T condition, producing the over-tempering of the original martensite, which generates a soft zone.

Figure 9.16 shows the microstructure evolution for the different zones formed during LW of AISI 4130 in annealed condition.

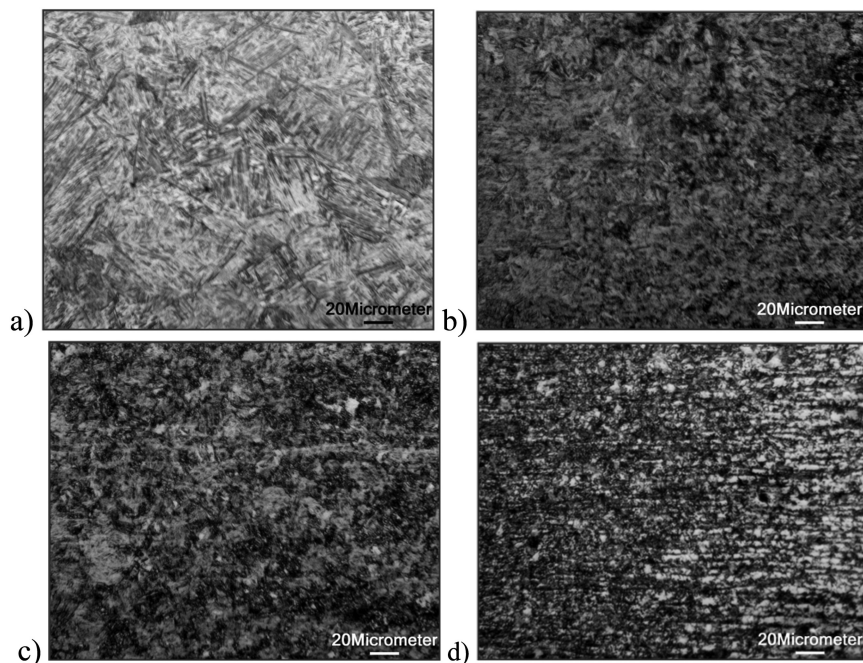


Figure 9.16 a) WM, b) CGHAZ, c) FG/ICHAZ and d) SCHAZ/BM for AISI 4130 in annealed condition.

The narrowness of it depends on the heat input (Nascimento 2010). Tsay et al. (1992) have reported that the LW process is the most employed process to reduce HAZ softening. This aspect is relevant due to its effect on the final mechanical properties of the component, especially in the SRM case performance.

Pre-heating the material before welding could reduce the hardness and brittleness of the welded joint, due to bainite formation (Mensur et al. 2020). However, the mechanical properties achieved do not meet the requirement for SRM case manufacturing.

Post-Weld Heat Treatment (PWHT)

As was previously mentioned, welded joints of AISI 4130 steel need a PWHT in order to reduce the hardness of the fresh martensite formed, to adjust the mechanical properties. Therefore, the PWHT usually consists of a tempering to temperatures in the range of 400–550 °C, during 1–2 hs. Starting with the material in a Q&T condition previous to laser welding, Souza Neto et al. (2015) have obtained a hardness of 450 HV after tempering at 400 °C per 2 hs, producing the tempering of the fresh martensite formed at WM, CG-HAZ and

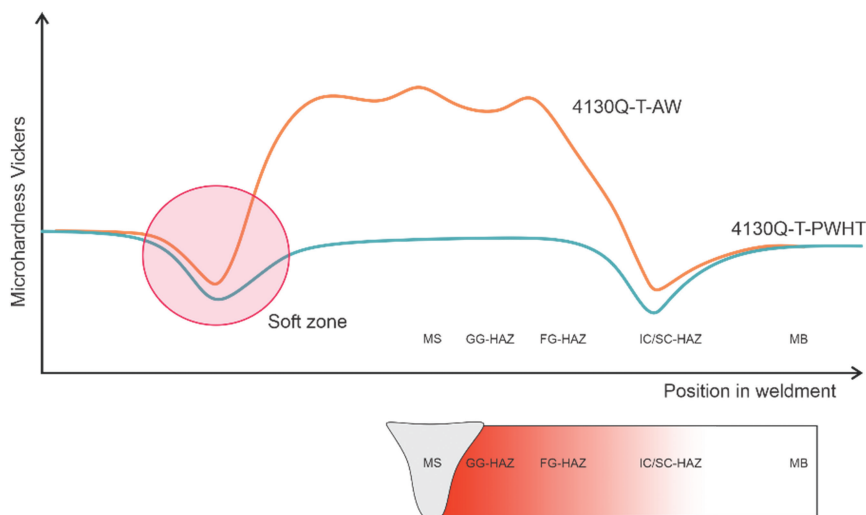


Figure 9.17 Scheme of microhardness profiles along the laser welded joint of AISI 4130 (Q&T) in AW and PWHT conditions.

FG-HAZ. Li et al. (2011) have simulated a welding cycle in AISI 4130 with an initial Q&T condition, reporting that after PWHT the hardness decreased in the WM and high temperature HAZ according to the experimental results.

After PWHT the IC/SC-HAZ remains soft due to the presence of ferrite and carbides, formed during the weld thermal cycle, but an additional softening usually occurs. Tsay et al. (1992) have investigated the effect of PWHT temperature for welded AISI 4130 in Q&T condition, and the temperatures studied were 200 °C, 300 °C, 400 °C and 525 °C for 1 h. They have reported a decrease in hardness with the increase in PWHT temperature, as well as the increase of the over-tempered zone size (SC-HAZ). In Figure 9.17 a scheme of the microhardness profiles obtained in AW and PWHT conditions of a laser welded joint of AISI 4130 in Q&T condition can be seen.

If the welding is done with the material annealed and Q&T is applied after welding, the resultant microstructure should be uniform and composed of tempered martensite, maintaining only marginal characteristics of the original welding zones.

Maraging Steel

As Welded Condition

The thermal cycle produced during laser welding of MA steel in SRM case fabrication also generates a microstructural gradient from the weld centreline

to the BM, associated with the peak temperatures and cooling rates established in each point. The thermal fields produced in the welded part are controlled for the welding parameters used, mainly laser energy density (E_d) (Lang and Kenyon 1971).

Weldability is one of the most important characteristics of the MA steels. Regardless of the grade and type of MA steel, the weldment zones have generally the same characteristics. MA are usually welded in annealed condition, followed by a low temperature PWHT. However, as it was previously mentioned, the steps of aerospace component manufacturing could lead to welding the MA steels both in annealed or aged condition.

The transformations that take place during welding in the volume of the material that reaches temperatures above A_{c3} are independent of the original condition of the BM, due to the dissolution of any precipitates and the formation of a full austenitic structure at the temperatures mentioned (Gupta and Raja 2020).

In the case of WM, due to the high alloying element content the solidification will occur in an unstable front, typically cellular-dendritic, producing a microsegregation phenomenon. On further cooling in solid state, the austenite transforms to martensite. Finally, the WM presents a martensitic microstructure with a solidification pattern, characterized by microsegregated zones, rich in Ni, Ti and Mo (Rohit et al. 2017). Fanton et al. (2014) according to Shamantha et al. (2000) have mentioned that the massive or lath martensite is relatively soft with a high dislocation density. Figure 9.18 shows the microsegregated martensite in a cellular-dendritic solidification pattern obtained in the WM of a laser welded joint of a C250 grade.

Four different regions can be identified in the HAZ of MA steel weldment: CGHAZ, FGHAZ, ICHAZ, SCHAZ. As was previously mentioned, the CGHAZ experiences temperatures high enough to cause full austenitic

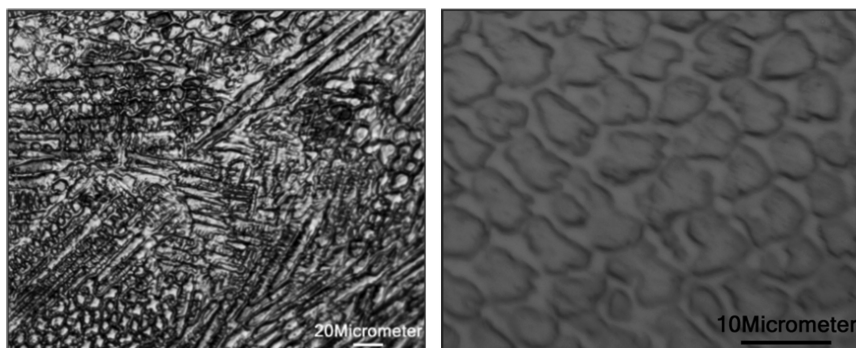


Figure 9.18 WM microstructure of a C250 LW in annealing condition.

transformation and grain growth. In the FGHAZ, the peak temperatures were close above A_{C1} and then the austenite grains formed cannot grow, experiencing a grain refinement (Reddy et al. 2009). The austenite of these regions transforms on cooling to room temperature to soft and dislocated lath martensite with the same characteristics of the original one in the annealed condition. CG-HAZ and FG-HAZ are both usually called Light Etched Zone, considering them as a single area (Gupta and Raja 2020). The obtained microstructure of these zones is independent of the BM starting condition. In Figure 9.19 the CG HAZ and FGHAZ of a laser welded C250 MA steel can be seen.

In the ICHAZ, the material is reheated to temperatures between A_{C3} and A_{C1} , causing the partial decomposition of the martensite into austenite. For higher temperatures up to A_{C3} , the martensite to austenite formation is controlled by a non-diffusional shearing mechanism (Gomes de Carvalho et al. 2013). For lower temperatures, a solid-state diffusion-controlled decomposition of original martensite in a depleted martensite (α') and enriched austenite (γ') takes place. This reverted austenite will not transform to martensite at subsequent cooling to room temperature. Some authors have reported that the microstructure obtained for this zone is a martensite matrix with a very low volume fraction of fine dispersed reverted austenite (Lang and Kenyon 1971). This zone is usually called Dark Zone, since due to the precipitation of reverted austenite and other phases it is observed dark when the microstructure is revealed with a convenient reagent (Shamantha et al. 2000). Gupta and Raja (2020) have also reported the presence of reverted austenite in the dark zone. The amount of reverted austenite at ICHAZ decreases with the decreasing heat input and with the increasing heating rate (Lang and Kenyon 1971). In this sense, LW will promote a lower reverted austenite in the dark zone.

If the material is welded in annealed condition, a fraction of aged martensite will also be present in this zone. If the material is welded in an aged

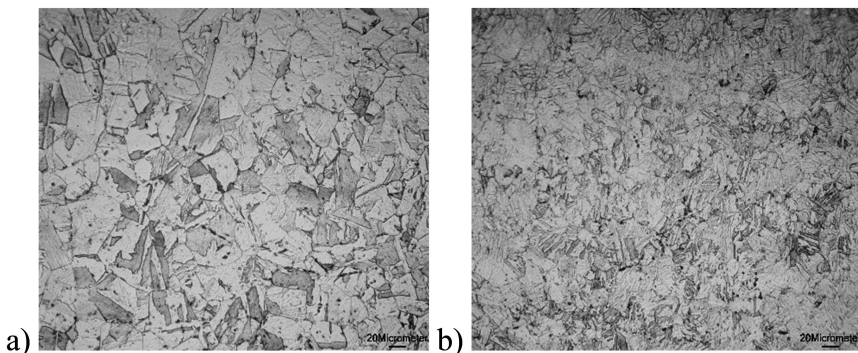


Figure 9.19 a) CGHAZ, b) FGHAZ for laser welded C250.

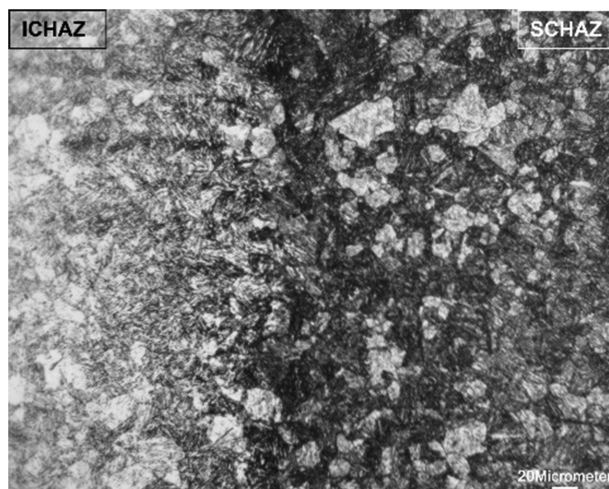


Figure 9.20. ICHAZ and SCHAZ microstructure of a laser welded C250 in annealed condition.

condition, the fraction of remaining martensite will be overaged (Lang and Kenyon 1971).

Finally, previously to reach the BM, there is the SCHAZ which has been reheated below A_{C1} temperature and presents different characteristics for each starting material condition (Lang and Kenyon 1971; Duran et al. 2021). Although commonly the peak hardening is achieved after long terms (several hours for this temperature range), the Co content of the alloy reduces the precipitation reaction time enough to cause slight precipitation in the IC/SCHAZ during the welding cycle of an MA steel in the annealed condition (Gupta and Raja 2020).

Figure 9.20 shows the microstructure at ICHAZ and SCHAZ of a C250 laser welded sample.

For the case of welds of MA steels in aged condition, the SCHAZ is characterized by a narrow soft zone of coarsened precipitates, which could produce a slight local softening. However, for both starting material conditions, when the material is reheated to temperatures where the reversion reaction tendency is high (close to A_{C1}), the austenite formation will take place (Shamantha et al. 2000; Sha and Guo 2009). In the case of aged material, the coarsening of the precipitates due to high temperature may collaborate with the formation of reverted austenite (Sha and Guo 2009).

Post-Weld Heat Treatment

The main reaction that takes place during PWHT is the ageing responsible for the hardening of the material, considering the material was welded in

the annealed condition. The martensite matrix of the WM, HAZ and BM is hardened by the precipitation of intermetallic compounds such as $\text{Ni}_3(\text{Mo}, \text{Ti})$ (Dehgahi et al. 2021) as was previously described for the BM. For T250 the precipitate size increases slowly for PWHT temperatures from 420 to 520 °C, but its volume fraction increases rapidly, reaching the peak hardening. For higher temperatures the overageing takes place (Li et al. 2016).

Additionally, regardless of the initial condition of the material, during PWHT in the WM the reversion of martensite will occur. Li et al. (2016) have reported the presence of reverted austenite in the WM of a T250 with PWHT at different temperatures (420–640 °C per 3 hs). This reaction is promoted by the strong segregation of Ni, Mo and Ti at the last liquid to solidify (intercellular/dendritic spaces) which causes the decrease of the martensite reversion temperature. This segregation was measured, showing an increase of segregation with PWHT temperature. For ageing temperatures from 420 to 520 °C the amount of austenite increases with the temperature because the driving force to martensite-austenite reversion also increases. Reddy et al. (2009) also have reported reverted austenite in the WM after PWHT at 480 °C per 3 hs in C250 steel. Due to the segregation in the WM, after PWHT the microstructure is composed by white islands of reverted austenite, darkened regions of martensite around the austenite islands and an aged martensite matrix (Lang and Kenyon 1971). The amount of reverted austenite formed is usually up to 16% in volume (Reddy et al. 2009; Duran et al. 2021). Figure 9.21 shows the microstructure of the WM of LW C250 with PWHT of 480 °C – 3hs. The solidification substructure and the reverted austenite formed at intercellular spaces of the aged martensite can be seen.

When the aged temperature exceeds 560 °C the authors have reported that new reverted austenite appears uniformly distributed within the martensite matrix in the WM and BM because the temperature of austenite equilibrium has been reached (Li et al. 2016).

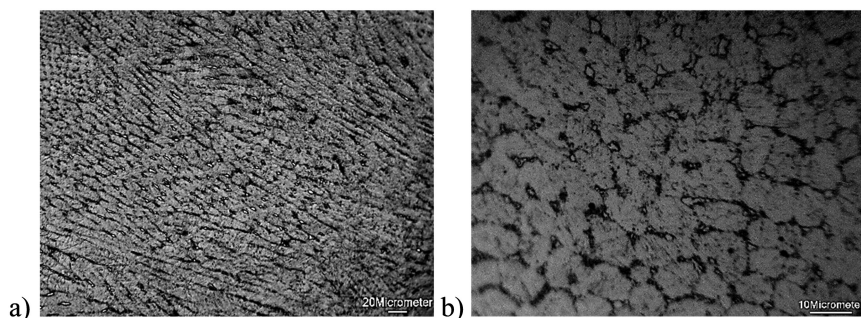


Figure 9.21 PWHT microstructure at WM of C250 LW aged at 480 °C–3 hs.

The CG-HAZ and FG-HAZ after PWHT have a microstructure of aged martensite maintaining the grain structure of the AW condition (Figure 9.22).

Reddy et al. (2009) have mentioned that the reverted austenite present in the ICHAZ remains unaffected after PWHT. However, it could be removed through a homogenized treatment at 1150 °C-1h, followed by a solutionized treatment at 850 °C-1 h and a final aged 480 °C-3 hs. Therefore, the hardness across the weldment attains values equivalent to the BM in aged condition.

Murthy et al. (2019) have compared GTAW vs. LW processes, concluding that after PWHT in both cases a slight hardness reduction is observed at ICHAZ which could be attributed to the presence of reverted austenite. However, there is a lower degradation for LW samples. Furthermore, they also stated that after different PWHTs (480 °C–510 °C per 3 hs) the reverted austenite in the ICHAZ remains unaffected, due to the presence of soft zone.

Figure 9.23 presents the microhardness profiles obtained on the mid-thickness line through the whole laser welded joint of C250 (annealed) in AW and two different PWHT conditions (480 °C-3hs and 520 °C-3hs).

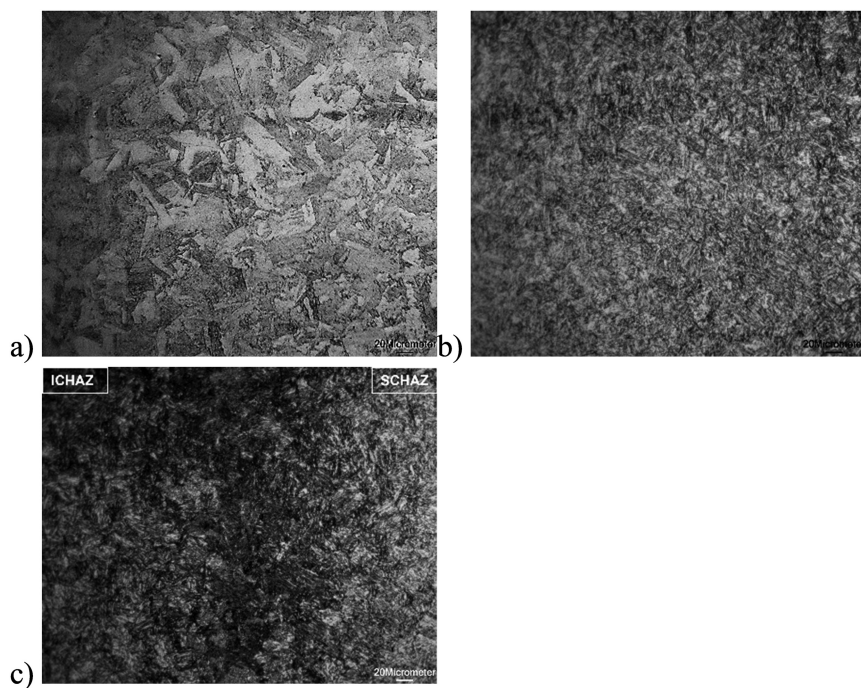


Figure 9.22 a) GG-HAZ, b) GF-HAZ and c) IC/SC-HAZ - PWHT 520 °C-3hs for C250 steel.

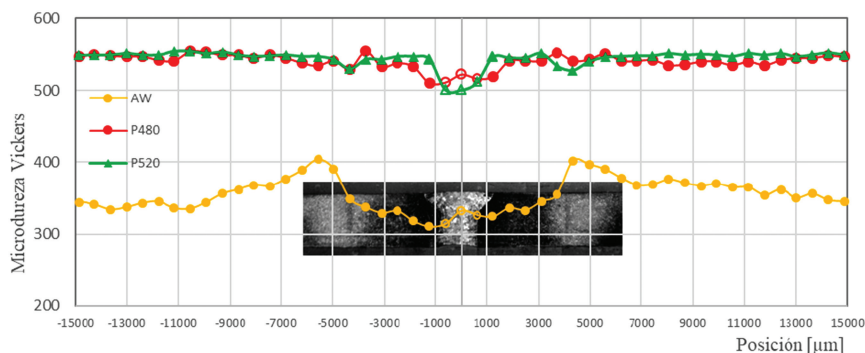


Figure 9.23 Microhardness profiles of LW C250 (annealed) in AW, PWHT-480 °C and PWHT-520 °C (Duran et al. 2022).

In the AW condition it is observed that an increase in the microhardness in the SCHAZ is related to the ageing of the martensite. The peak of hardness value (400 HV) is located near the A_{C1} temperature and then in the ICHAZ the hardness decreases due to the transformation of austenite to new soft martensite during cooling. When A_{C3} is reached the microstructure is completely composed of soft martensite, and the hardness value is the same as the original BM. This value is almost constant in FGHAZ, CGHAZ and WM (Duran et al. 2022; Lang and Kenyon 1971). For PWHT the hardness increased up to a uniform value associated with the ageing temperature and time (550 HV), except in the WM where the hardness is slightly lower (500–520 HV) also depending on PWHT time and temperature. This is related to the reverted austenite formed in that region. A somewhat lower hardness is also measured at SC/ICHAZ limit, where a fine dispersion of reverted austenite could explain it. In this case a significant variation of microhardness profile with the PWHT temperature (480 and 520 °C) is not observed.

In Figure 9.24 the microhardness profiles of the laser welded sample of C250 (aged) in AW, PWHT-480 °C and PWHT-520 °C can be seen (Duran et al. 2022).

If the material is welded in aged condition, a microhardness decrease was observed from the SCHAZ due to the overageing of the original microstructure. In the ICHAZ the hardness decrease is higher due to the additional formation of austenite that transforms into soft martensite on cooling to room temperature. In the FGHAZ, CGHAZ and WM the microhardness reached the minimum value of whole welded joint due to the fully soft martensite microstructure. With the PWHT the microhardness increased to a uniform value (550–570 HV) related with time and temperature applied, except in the WM, where it is something lower due to the presence of reverted austenite

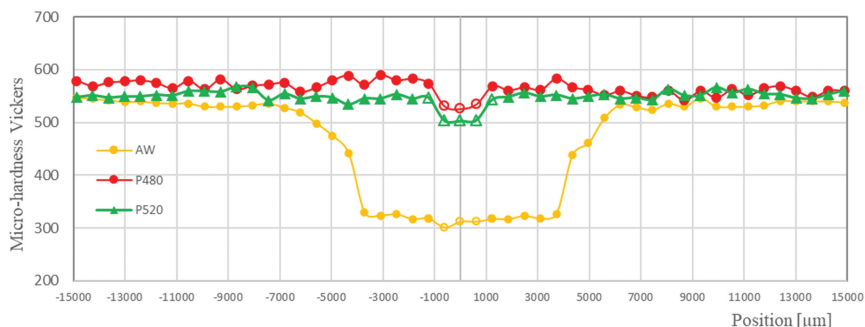


Figure 9.24 Microhardness profiles of C250 LW (aged) in AW, PWHT-480 °C and PWHT-520 °C (Duran et al. 2022).

(500–530 HV) (Duran et al. 2022; Lang and Kenyon 1971). In this case for 520 °C the hardness achieved is slightly higher than for 480 °C (Duran et al. 2022).

Mechanical Properties of Laser Welded Joints

AISI 4130

As Welded Condition

AW condition is an intermediate situation, before the final PWHT is required. Nevertheless, the analysis of mechanical properties in this stage allows to understand the evolution of it.

The hard and brittle martensite formed in the WM and high temperature HAZ after welding are responsible for the global mechanical behaviour of the welded joint.

Li et al. (2010) and Souza Neto et al. (2015) stated that the mechanical properties in the welded joint differ with respect to the BM, showing a decrease in the toughness and ductility. Mansur et al. (2020) have reported that in comparison with the arc welding process, the LW could generate higher temperature variations and higher cooling rates. In this context, the tendency for cracking increases due to the residual stress after welding, but this tendency could be reduced by tempering the material immediately after welding. Tsay et al. (1992) reported for an AISI 4130 welded in Q&T condition, the fracture mechanism in tensile test was intergranular along the columnar grains of the WM. They also have pointed out that the WM is a potential weak point in the weldment due to TME (*Tempering Martensite Embrittlement*) that reduces the impact toughness and increases the crack growth rate, causing a potential

intergranular failure. It is important to note that after welding residual tensile stresses at the welded joint could affect the toughness of the component. Joshi et al. (2016) have reported that the fracture of a tensile test sample, extracted from a welded joint of an AISI 4130 welded in Q&T condition, was located in the ICHAZ, which is the minimum microhardness zone. Furthermore, the same authors have reported that after LW the tensile properties were superior to GTAW, associating this result with the drastic reduction of the heat input in the first one. The joint efficiency achieved by LW process for an AISI 4130 welded in Q&T condition was above 90% as compared to GTAW process where the joint efficiency was less than 70% (Joshi, 2016). Tsay et al. (1992) have reported for AISI 4130 laser welded in Q&T (860 °C 0.5 hs, oil quenched and tempered at 525 °C – 1h) a UTS strength greater than 1000 MPa in AW condition. The fracture toughness was low ($< 40 \text{ MPa } \sqrt{m}$) due to the coarse columnar structure and untampered martensite in the WM and high temperature HAZ.

Post-Weld Heat Treatment

PWHT for AISI 4130 is required in order to improve the mechanical properties of weldments by tempering the fresh martensite formed in WM and HAZ. Usually, the temperatures employed are in the range of 400– 550 °C, as was mentioned for Q&T heat treatment.

Souza Neto et al. (2015) have reported that after PWHT the ductility is recovered through the tempering (400 °C – 2 hs) of martensite attaining tensile strength near to 1000 MPa. Although the tensile values reached for LW and GTAW are similar to those obtained in BM, Souza Neto et al. (2015) have mentioned that for LW process, the material recovered the ductility practically equal to the BM, while for GTAW the ductility recovery was lower.

Li et al. (2011) have simulated a welding cycle on an AISI 4130 in Q-T condition. They reported that after PWHT (640 °C -1.5 hs) the fracture toughness in brittle zones has been improved due to the martensite tempering. The fracture surface was composed mostly of elongated dimples. However, the weldment strength is slightly lower than the material strength in Q&T condition due to the presence of bainite in the CGHAZ and ferrite in the IC/SCHAZ. Tsay et al. (1992) have investigated the effect of PWHT temperature for an AISI 4130 welded in Q&T condition. The temperatures analysed were 200 °C, 300 °C, 400 °C and 525 °C for 1 h. They have reported that the hardness was decreasing with the increasing PWHT temperature. They have reported higher crack growth rates in the WM and HAZ after tempering at 300 °C due to the Tempering Martensite Embrittlement (TME) which has not been completely eliminated. For the weldments treated above 300 °C the crack growth rates decrease considerably. Although the over-tempered zone (SCHAZ) is narrow,

it widens and softens when the PWHT temperature increased. The microstructural variation directly impacts the mechanical properties of weldments, with a ductility increase and strength decrease as a result of PWHT temperature rise (Tsay 1992). Mansur et al. (2020) have reported that immediately after tempering at 530 °C per 2 hs, the residual stress is reduced by the recovery mechanism.

Generally, the best performance is achieved for PWHT temperatures in the range of 440–520 °C, reaching tensile strength higher than 1000 MPa with relatively good ductility and toughness.

Maraging Steel

As Welded Condition

As it was mentioned before for MA steels the AW condition is also an intermediate situation that enables to better understand the results obtained in PWHT final condition.

Reddy et al. (2009) have tested maraging C250 weldments in annealed condition by tensile test. They have reported that after the LW process the strength achieved was around 1000 MPa and the failure location was away from the WM; it was located in the ICHAZ, exhibiting a ductile fracture. The authors have attributed that to the presence of reverted austenite in the dark zone, which is soft and ductile. In relation to the toughness, they have pointed out that it depends on the volume fraction of the second phases, inclusion and cell sizes. For the AW condition, they have reported a decrease in the impact toughness from 110 J/cm² to 80 J/cm² in relation to the base material in annealed condition due to the cellular structure in the WM. Li et al. (2016) have pointed out that T250 in AW condition is formed by martensite lath without precipitates which leads to a low strength and high toughness. The authors have proposed that the static toughness is directly proportional to the dimples size, measured at the fracture surface of the tensile tested sample. They have reported the strength and static toughness for both BM and welded joint in different conditions (AW and PWHT). For AW condition, the results were similar than the ones obtained for BM: almost 900 MPa for tensile strength and 90 MJm⁻³ for static toughness. These results would indicate LW did not introduce significant defects. Duran et al. (2022) have reported similar results for C250 laser welded in annealed condition and compared to the BM. The authors have achieved values close to 900 MPa. The fracture is located at the WM, due to its lower hardness. An et al. (2012) have reported that joint efficiency decreased from 90% to 70% after LW for a maraging steel in aged condition. Although the

HAZ is narrow, the material degrades completely achieving tensile strength close to 1300 MPa with a low ductility to 2%. The failure zone was localized in the HAZ previous to the BM, and they have been associated with a small fraction of reverted austenite, which makes the material weaker even than the WM.

Post-Weld Heat Treatment

After PWHT the weldments of maraging steel achieve mechanical properties very close to the material in aged condition. Li et al. (2016) have studied the mechanical properties of laser-weldment T250 after different PWHT (420 °C to 640 °C per 3 hs) and have associated the result with the hardening precipitation and reverted austenite. For the specimens treated from 420 to 520 °C per 3 hs the authors have reported a tensile strength increase until 1640 MPa with a toughness decrease to 40 MJm⁻³. For the weldments aged from 420 °C to 520 °C, the precipitate is the main contributor to strengthening because these interact with the dislocation present in the martensite matrix. At the fracture surfaces of tensile samples of weldments aged in this temperature range there was observed more dimples density. At temperatures above 520 °C the tensile strength starts to decrease until 1100 MPa while the toughness increases to 80 MJm⁻³ approximately. Li et al. (2016) have reported that initially the precipitates grow up and then dissolve back with the formation of reverted austenite in the matrix. The dimples in the fracture weldment surface become bigger and deeper when the ageing temperature exceeds 520 °C.

Reddy et al. (2009) have reported that after PWHT at 480 °C per 3 hs the weldments achieve a tensile strength to 1650 MPa failure in the ICHAZ. The authors have attributed that to the presence of the reverted austenite which is soft and ductile and the ICHAZ zone is weaker than the other weldment parts, even to the WM. The same authors have reported that after a homogenizer treatment at 1150 °C per 1h, followed by solutionized treatment at 850 °C per 1 h and aged 480 °C per 3 hs, the weldment achieved a tensile strength of 1750 MPa failure in the WM. They have pointed out that the presence of reverted austenite in the fusion zone reduces the toughness because the soft austenite will deform preferentially and might reach its critical strain for fracture at early stages setting up a series of micro-cracks in a harder matrix (Reddy et al. 2009). The toughness is improved when the amount of austenite in the weldment is reduced in the WM as well as in the ICHAZ after homogenizer, solutionized and aged treatment. Duran et al. (2022) have reported that after PWHT at 480 °C per 3 hs the same strength as the base material in the aged condition has been achieved (1600 MPa approximately); however, the

ductility of the weldments has been decreased in relation to the base material in the same condition.

Laser Metal Deposition (LMD) for SRM

Metal additive manufacturing (MAM) has grown exponentially in the last years driven in part by its favourable impact in the aerospace industry. The main goals of this technology are lead time reductions, use of new materials, unique design solutions (lightweight and highly efficient), consolidation of multiple components and elimination of traditional joining processes, among others. These opportunities have been commercially applied to several aerospace components, including liquid rocket engines, propellant tanks, turbomachinery, satellite components, etc. (Blakey-Milner et al. 2021; Freeman 2021).

Among MAM available techniques, DED is not limited in the size of the component to produce. Large liquid fuel engine components have been recently developed using DED. Entire rocket fuel tanks are possible to manufacture in a single piece or far fewer pieces by means of DED, with significant savings of material and fewer tooling requirements, reducing considerably the cost and lead time of these components (Svetlizky et al. 2021).

LMD is one of the more used DED techniques in which the material is deposited locally feeding it directly into the weld pool. The energy source to create it in this case is a laser. Since LMD does not use a powder bed, it is not limited by the build size of the component, and can therefore be manufactured with very large components. The dimensional resolution achieved in components built with LMD is lower (about 1 mm) than the one obtained with PBF techniques. In this sense, LMD is used for larger components which do not require high dimensional resolution (Blakey-Milner et al. 2021). Also, it can be beneficially used when post-processing operations could be performed.

The material feedstock could be in the form of powder or wire. Powder LMD (P-LMD) was first developed and is one of the most used. The deposition head could be attached to a robotic or gantry system and usually employs a local inert gas. Wire LMD (W-LMD) uses the same principles, but the feedstock material is a wire, which may be off-axis or co-axial. It has a higher material efficiency and lower cost, but lower resolution (Blakey-Milner et al. 2021). In the last years the use of wire has increased, due to its higher availability and lower costs. Figure 9.25 shows the main processing parameters of P-LMD (Svetlizky et al. 2021).

In P-LMD the metal powder is carried to the deposition zone by an inert gas. C300 grade is the most applied MA steel in AM field. In Figure 9.26 C300 MA powder and its particle size distribution can be observed, which shows an average size close to 60 microns.

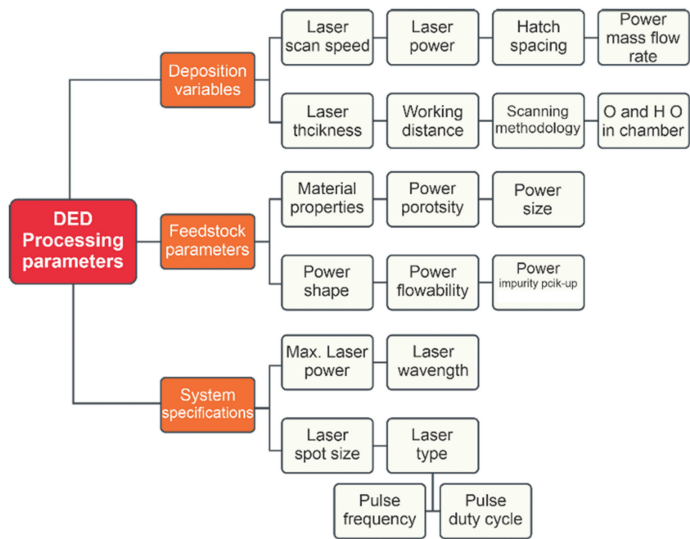


Figure 9.25 P-LMD process parameters.

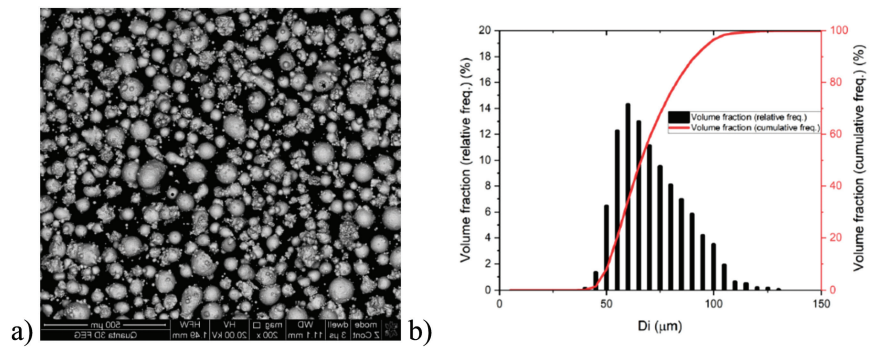


Figure 9.26 C300 powder: a) SEM of particles, b) particle size distribution (Ronda et al. 2022).

Some equipment for P-LMD have multiple nozzles for powder feeding, which are capable of producing products with different composition along it. Also, material mixtures can be produced (Blachowicz et al. 2021). LMD can also be used with filler material in the form of wire. In the last time, this alternative appears to be more convenient for several applications, considering the lower cost and more availability of the wire.

LMD process has many process parameters. The main ones are laser power, scan speed and powder feed rate. Ronda et al. (2022) studied the effect of powder feeding rate, which produces different deposited layer thickness, on

the microstructure and mechanical properties of C300 additive manufactured parts. They analysed three layer thicknesses: 0.5, 0.75 and 1 mm, maintaining the other parameters constant. After printing, the samples were subjected to annealing (840 °C, 2h) and ageing (480 °C, 5h). Porosity slightly increases with the layer thickness. Figure 9.27a shows the obtained microstructure for 0.5 mm. The overlapped weld beads as well as the layer thickness of 0.5 mm could be seen.

A cellular-dendritic solidification pattern was observed for all samples (Figure 9.27b). Lower layer thickness produced a finer microstructure due to a higher cooling rate. The microstructure is composed of lath martensite and a significant amount of austenite, which the authors define as retained (Figure 9.27c). Microhardness was not affected by the layer thickness (615 HV approx.), while the tensile properties were similar for 0.5 and 0.75 (σ_{UTS} =1940 MPa) but fall-down for 1 mm with brittle fracture. This last result was associated with the presence of unmelted particles and porosity (Ronda et al. 2022).

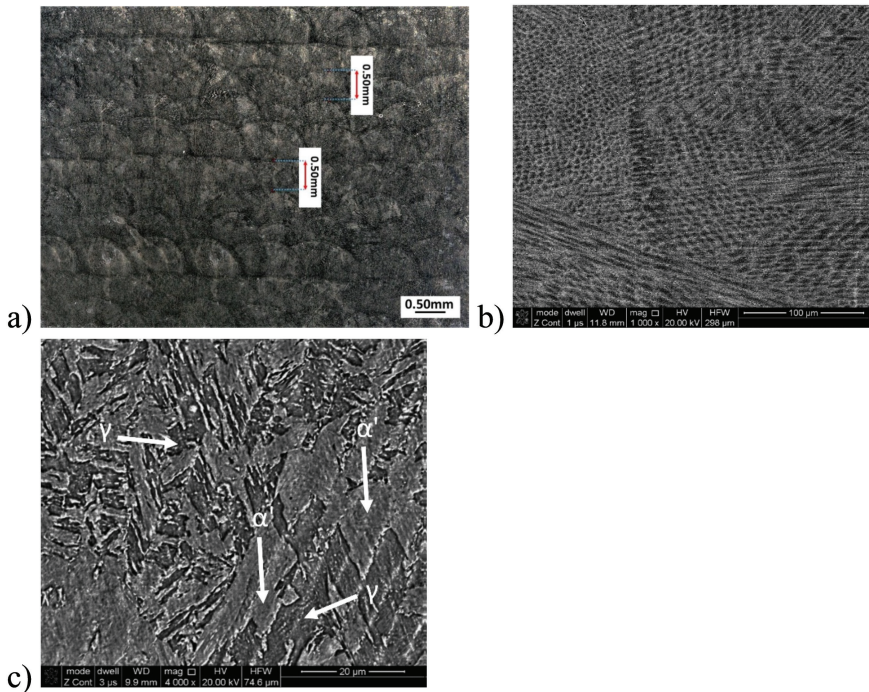


Figure 9.27 Microstructure of C300 LMD with 0.5 mm layer thickness in print condition. a) Overlapped weld beads, b) cellular-dendritic structure, c) lath martensite and austenite (Ronda et al. 2022).

Yao et al. (2018) have studied the effect of LMD parameters on C300 printed parts characteristics, using an Yb fibre laser. They analysed the effect of laser power (P), scanning speed (v) and powder feeding rate (fr) on micro-structure, defects and mechanical properties. Laser energy density ($Ed=Pv/d$) is also a relevant parameter, which is associated with P , v and spot diameter of the laser beam (d). They have shown that P was the most relevant parameter related to the tensile strength of the part in as print condition, followed by v and fr , reaching between 700 and 960 MPa of tensile strength depending on the manufacturing conditions. The fracture surface for the best condition was ductile with microvoid coalescence, while for the worst case a solidification pattern was observed on the fracture surface associated with low ductility values. Authors related these results to porosity level (density). A minimum energy density of 180 J/mm² ensured 100% of the printed part density (no pores formation). After manufacturing, they applied a solution heat treatment and then ageing at 490 °C between 1 and 10 hs. Tensile strength and ductility increased with ageing time, reaching approx. 1550 MPa and 12%, respectively, for 10 hs.

Grain size increased with increasing laser density, with the consequent reduction in hardness and other mechanical properties. It was also observed that the porosity decreases as the laser power increases, due to higher solidification times (Yao et al. 2018).

Tan et al. (2021) proposed that the main advantages of using LMD in these materials are related with: Intrinsic Heat Treatment (IHT), grain refinement and efficient method that matches the requirements of the geometrically complex aerospace products (low scale of production and excellent mechanical properties). IHT is associated with the effect that each deposited layer exerts on previously deposited material, generating an in-situ heat treatment with each new layer. IHT may produce the precipitation during the build-up of the product, without needing an additional heat treatment (Kürnsteiner et al. 2017). The high cooling rates associated with LMD process (up to 105 °K/s) (Wang et al. 2008) lead to grain refinement in the as-print parts, which improves its mechanical properties.

Li et al. (2022) analysed the effect of oxygen content in the atmosphere of processing in LMD of C300 MA steel in powder. They print walls in three atmosphere conditions: Air, 500 ppm and 50 ppm of oxygen. As oxygen content was decreased the surface oxidation and roughness were diminished, and the dimensional accuracy was improved. Also, the laser-induced plasma temperature was increased, due to the lower ionization potential of the oxygen. As a consequence, primary dendritic arm spacing was increased. Nevertheless, the microhardness and tensile properties were improved with the decrease in oxygen content. The authors associated this observation with the higher

ageing effect due to the mentioned higher temperatures reached and lower cooling rates. The fracture surfaces of samples printed with 50 ppm were ductile, while in the case of those printed in air, they showed oxides and produced brittleness.

When the porosity is controlled, the results reported are similar to those obtained for wrought materials, both in annealed and in aged conditions.

AISI 4130 has been also studied for its use in AM parts for aerospace applications (Tan et al. 2021) showing good results.

For both materials microhardness and tensile properties generally match or exceed the ones obtained from conventional manufacturing techniques (Blakey-Milner et al. 2021). Related to fracture toughness or fatigue response, aspects like porosity, residual stresses or surface roughness could affect the final properties, being usually lower compared with conventional processes (Blakey-Milner et al. 2021).

Post-processing is usually needed in LMD parts involving support removal, surface finishing and/or heat treatments. Generally, it must take care to prevent damages to the printed part. Surface polishing or machining could be needed, especially for matching surfaces. Heat treatments could be also applied to reduce residual stress field or to adjust the microstructure. For MA steels, solutionization and/or ageing are normally used (Blakey-Milner et al. 2021).

Concluding Remarks and Future Outlook

Aerospace industry has grown in the last years, and it is expected that it will continue. SRM case fabrication is not the exception and LW is part of the present and it is believed also the near future.

AISI 4130 has shown during years its applicability as SRM case material. Nevertheless, its moderate weldability and the difficulties to apply the heat treatment after welding limit its application. Maraging steels has been established as the main material used for SRM case fabrication. New grades of C450 (>3000 MPa) are under development and the resultant fracture toughness is the key challenge. Additionally, it has become relevant to develop cobalt-free maraging steels with improved properties, close to the higher strength C grades. This will reduce the cost of MA steels and save a strategic material such as cobalt.

Laser Welding is a mature technology for welding AISI 4130 and MA steels in SRM construction, presenting several advantages in comparison to other welding techniques. In particular MA steels will be the most used materials in the near future for this application, being the newer grades a challenge to LW.

The precipitation mechanisms during ageing in PWHT are well known. The final mechanical properties are affected by the reverted austenite formation in WM or HAZ. The observation of reverted austenite in the dark zone is a topic that has not been completely resolved. There are only a few old images to show the presence of this phase, even with the advanced techniques available today to characterize materials. This aspect is an issue where more work is needed.

Fracture toughness of laser welds of the higher strength grades MA steels will be the more relevant aspect to analyse.

AM has been positioned as a key technology. Particularly, LMD for aerospace industry have now several applications running and a high potential for many other applications in this industry. Pressure vessels for liquid rocket engines have been constructed, as well as nozzles and many other rocket parts.

The quality controls related to aerospace manufacturing are very rigorous. This is a key aspect related to AM process, due to their complexity and numerous manufacturing parameters that could influence the component quality. Many studies have reported a high dispersion of the mechanical properties of AM parts, especially in fatigue or fracture toughness. Several standards are under development to address this issue; considering that with appropriate quality controls and optimized procedures, it could be minimized. Structural integrity is a very important issue related to the aerospace applications. It is still a field where more work is needed related to the performance of LMD parts of MA steels.

Despite these considerations, SRM case is an application for LMD with several advantages, and it is deemed that LMD will increase in aerospace applications, also as an available technology for SRM case manufacturing.

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Chapter 10

Laser Welding for Micro Process Devices

T. Gietzelt, M. Mahmoudizahdeh,

T. Wunsch and R. Dittmeyer

Institute of Micro Process Engineering (IMVT), Karlsruhe Institute of Technology (KIT),
Po. Box 3640, 76021 Karlsruhe, Germany

Abstract

Laser welding is a thermal joining technology, which has a narrow, local heat-affected zone at rather high welding velocities. Moreover, no filler material is required, although a very good dimensional accuracy of the parts to be joined is needed. In this regard, adapted clamping devices should be used. The cross-section of the weld seam can be widely varied and adapted depending on the applications. At high welding velocities, the inlet of the laser weld narrows as the effect of the heat conductivity of the material decreases. When joining multilayer stacks, bulging of the weld seam at the surface may be challenging. For through welding, auxiliary trenches are required to mask the bulging of the weld seams and to weld additional layers on top. Moreover, for consecutively circumferential welding of multilayer stacks, gaps arise between adjacent layers. Adapted strategies in terms of welding sequence, welding direction, and welding parameters are required to bridge it. If several layers are to be through-welded, e.g. for integration of thin functional layers, further issues arise: Each gap acts as a thermal insulation. The quality of the clamping of stacks significantly affects the welding result. Gaps or thermally insulating intermediate layers may significantly affect welding penetration depth. Welding of micro process devices may suffer from locally different heat dissipation conditions, altering welding results. It is recommended to perform preliminary tests to evaluate welding conditions and to optimize welding parameters. By pulsed laser welding, the energy by unit of time

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and unit of length can be adopted to prevent distortion of thin parts. However, materials with a large solidification range may be subjected to hot cracking due to the remelting process.

Keywords: Laser welding, Local heat-affected zone, Micro process device, Weld seam cross-section, Multilayer stack

Introduction

Laser welding is a very well-suited joining technology for micro process devices due to high welding velocities and a low heat-affected zone. It is completely different to other techniques used for joining of multilayer parts, e.g. diffusion bonding. Laser welding is applicable to all sorts of metals and high-melting alloys. Nevertheless, reactive metals such as titanium and tantalum require a shielding gas to avoid the formation of brittle phases.

For multilayer assemblies containing mechanical microstructures, the weld seam geometry may be affected by locally different conditions of heat dissipation, a point that is neglected in the macroscopic world. For through welding, the transition to the next layer may act as thermal insulation, affecting welding penetration depth. Also for integration of thin membranes consisting of different materials, additional effects may arise. According to the binary phase diagrams, brittle phases may occur. However, at high welding speeds, no thermodynamic equilibrium is reached, and in some cases formation of brittle phases can be avoided.

In general, laser welding could be operated in two different modes: continuous wave (cw) and pulsed mode. The geometry of the weld seam cross-section can be tailored in a wide range according to the specific application. For cw-mode, an energy density of approximately 1 MW/cm^2 is required to form a metal plasma keyhole, enabling deep penetration welding. The heat input per time is very high, which may cause internal stresses and thermal distortion. At high energy densities, very high welding speeds of several meters per minute could be realized. The heat-affected zone is very limited or almost negligible. By pulsed laser welding, the energy input per unit of length or per unit of time can be adapted. Hereby thermal distortion of assemblies made of thin sheets, possessing a low thermal mass for heat dissipation, could be reduced. However, to achieve a vacuum gas tightness, the single pulses must overlap by approximately 75%. Therefore, some material is remolten multiple times, which might, in turn, cause crack formation.

The susceptibility to hot cracking during laser welding depends strongly on the solidification interval of alloys: A wide solidification range causes

formation of hot cracks at grain boundaries, where the remaining melt is located. Decreasing the temperature gradients by pre- or post-heating could help to reduce the effect. Austenitic stainless steels and nickel base alloys in solution-annealed condition can be welded very well, contrary to most aluminum alloys.

The interaction of the laser radiation with the metal may be different depending on the type of material and the wavelength used. For example, copper shows a very poor infrared light absorption around 1 μm wavelength but depends strongly on temperature as well. If a melt is formed, it increases drastically, leading to a gradient of the weld seam width with length. This effect is also supported by a high heat conductivity and depends strongly on welding velocity. To overcome this issue, frequency-doubled or frequency-tripled lasers should be used for copper.

Materials and Methods

Impact of Welding Parameters

There are many process parameters which should be considered together with technological parameters, and that are interact to each other. The main process parameters, however, are laser power, welding velocity, and focal position, which are discussed in the following section.

The focal position, for example, is a process parameter that could change during operation and that is related to the accuracy of the work piece position regarding the lens position. The impact on the energy density at the surface level and the position of the maximum energy density with respect to the work piece surface, respectively, depend strongly on the focal length of the lens. In case of solid-state lasers, it depends on the diameter of the light conducting cable (lcc) used (Figure 10.1).

From Table 10.1 it can be seen, that for a 100 μm -lcc the energy density at the surface level is much more affected by the focal position than that of 400 μm -lcc. Consequently, also the welding results differ much more for a 100 μm -lcc: It may cause solidification at the surface level first, while melt remains below, leading to pore formation by thermal shrinkage.

The impact of the welding velocity for cw-welding on the weld seam cross-section and welding penetration depth is also affected by the heat conductivity of the material. For austenitic stainless steel (heat conductivity 15 W/m $^{\circ}\text{K}$), the funnel-shaped inlet area decreases significantly starting at a welding velocity of around 5 m/min. Figure 10.2 shows that for 400 μm -lcc, due to lower energy density, the welding penetration depth decreases significantly with increase in

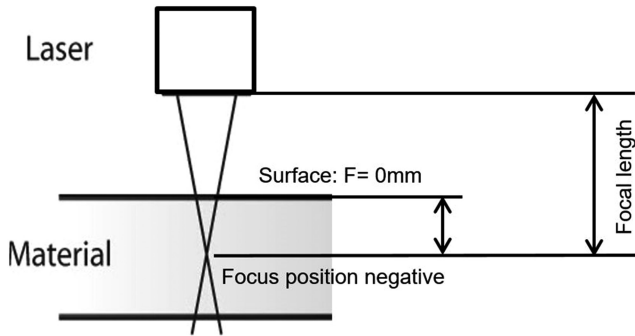


Figure 10.1 Scheme of the optical condition of focal length and focal position related to the surface level of the work piece.

speed from 1 to 5 m/min. For smaller lcc-diameter, this effect is shown with a less change in the penetration depth.

The interaction of welding velocity and energy density depending on lcc-diameters is even higher for materials possessing a higher heat conductivity, e.g. aluminum and copper. For aluminum alloys, it must be kept in mind that already small amounts of alloying elements have a huge impact on thermal conductivity: Whereas it is 300 W/m²K for pure aluminum, the heat conductivity is decreased to about 160 W/m²K for AlMg3 (AW-5754).

Technological Impacts on Laser Welding

The focal position is not only an important parameter for laser welding, but also it is subjected to technological constraints in terms of positioning accuracy of the work piece. The most critical impact of the focal position on the laser weld seam, however, is defined by the focal length of the lens. It strongly affects the change in energy density with focal position. Long focal lengths are preferred to limit this effect.

Laser welding is based on the interaction of laser radiation with solids, generating heat. No interaction takes place with air. Hence, when circumferential welding multilayer assemblies where shrinkage adds up between individual layers at the rim, welding results may vary. This is in close relation to the lcc-diameter used, as illustrated in Figure 10.3 and Table 10.2.

For the calculation of coupled laser power, a rectangular cross-section of the gap was assumed, which overestimates the reduction for larger gaps at small lcc-diameters. However, it should be noted that the impact of the technologically defined gaps has considerably more effect for small lcc-diameters

Table 10.1 Impact of the focal position on the energy density at surface level for a focal length of $f = 150$ mm, beam outlet diameter at collimation 33 mm for light conducting cables 0.1 and 0.4 mm in diameter, respectively.

Focal position related to surface level [mm]	lcc-diameter = 0.1 mm				lcc-diameter = 0.4 mm			
	Diameter at surface level [mm]	Cross-section [mm ²]	Energy density for P= 3 kW [MW/cm ²]	Decrease in relation to previous [%]	Diameter at surface level [mm]	Cross-section [mm ²]	Energy density for P= 3 kW [MW/cm ²]	Decrease in relation to previous [%]
0	0.1	0.007854	38.197		0.4	0.125664	2.387	
-1	0.32	0.080425	3.73	-90.2	0.62	0.301907	0.994	-58.4
-2	0.54	0.229022	1.31	-64.9	0.84	0.554177	0.541	-45.6
-3	0.76	0.453646	0.661	-49.5	1.06	0.882473	0.34	-37.2

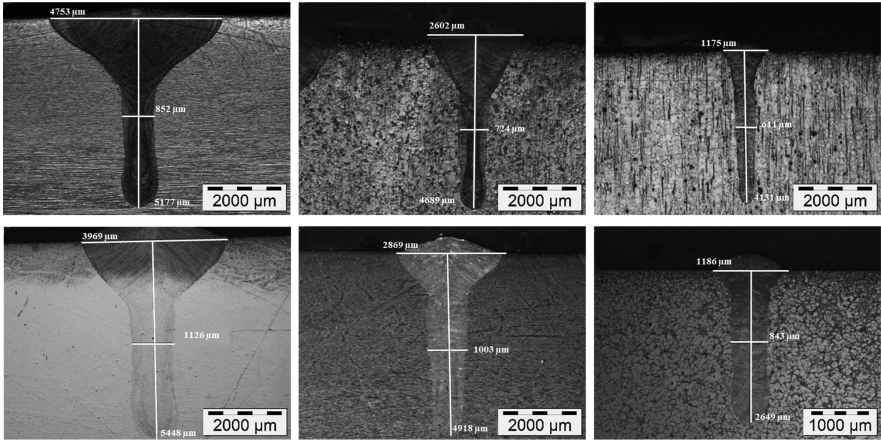


Figure 10.2 Impact of the welding velocity on the weld cross-section on stainless steel AISI 304 (1.4301). Focal length 150 mm, P = 3 kW, focal position = -1 mm. Top: lcc-diameter 0.1 mm, below: lcc-diameter 0.4 mm. From left to right: v = 1; 2; 5 m/min.

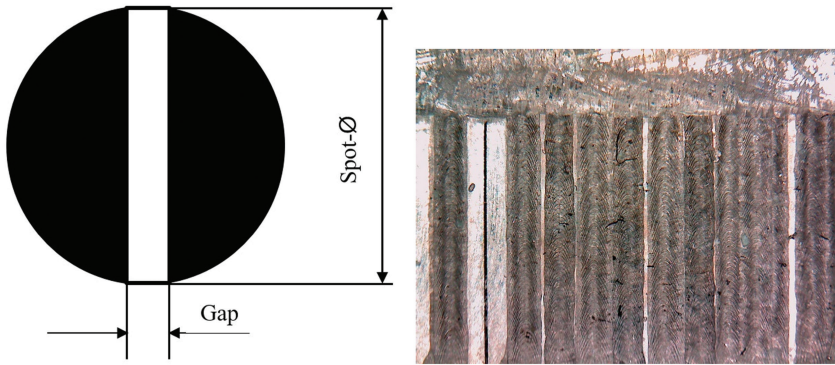


Figure 10.3 Left: Relation between the lcc-diameter and gap on laser power coupling. Right: Gap formation for consecutive circumferential laser welding of multilayer stacks, compared to the laser welds for joining the individual layers in Figure 10.9, left.

than for larger ones. On the other hand, small lcc-diameters provide a stronger impact on the weld cross-section, e.g. when varying the focal point position, as depicted in Table 10.1. For large lcc-diameters, however, the welding process is more stable and reacts less strongly to process deviations.

For materials, possessing high heat conductivities, and for small parts with a low thermal mass, there might be a positive correlation between the weld seam lengths and weld seam width.

Table 10.2 Percentage of coupled laser power depending on gap width for different lcc-diameters.

lcc-Ø [μm] →		50	100	150	200	400	600
Gap [μm] ↓	10	74.54	87.27	91.51	93.63	96.82	97.88
	20	49.07	74.54	83.02	87.27	93.63	95.76
	30	23.61	61.80	74.54	80.90	90.45	93.63
	50	n.a.	36.34	57.56	68.17	84.08	89.39
	100	n.a.	n.a.	15.12	36.34	68.17	78.78
	200	n.a.	n.a.	n.a.	n.a.	36.34	57.56

Moreover, there is a correlation of the absorption of the laser radiation depending on the wavelength used and the material to be welded. Regardless of a very high heat conductivity, copper also shows a low absorption of about 5% for infrared lasers around 1 μm wavelength (Haubold, Ganser, Eder, & Zäh, 2018; Sa et al., 2018; Sadeghian & Iqbal, 2022; Zediker, Fritz, Finuf, & Pelaprat, 2020). However, it increases with temperature and jumps when melting. Therefore, the welding result changes with weld seam length (Chen, Bi, Nai, & Wei, 2015). To prevent damage to the laser by back reflection, the optics must be tilted by 2–3° from a vertical position. Better results for copper are obtained for frequency doubled (green) or tripled (blue) lasers, offering higher absorption.

Material Scientific Considerations: Susceptibility to Hot Crack Formation

Susceptibility to hot cracking is mainly caused by the width of the solidification range of an alloy. During solidification, grains start to grow according to the thermodynamic equilibrium. With decreasing temperature, the percentage of the remaining melt between grains decreases, superimposed by thermal shrinkage of already solidified material. Eventually, a thin melt seam remains between the grains that cannot bridge the thermal shrinkage, and hot cracking occurs (Figure 10.4). For laser welding, this effect is superimposed by nonequilibrium high solidification rates. This is particularly evident for pulsed laser welding. Therefore, crack-free seams are more likely achieved by cw-laser welding.

On the other hand, internal stresses during solidification could be released elastically over short distances. For example, hypereutectic Al-Si alloys can effectively relieve shrinkage stresses by precipitation of primary silicon, finely dispersed at short distances inside grains (Figure 10.5, right). Hypereutectic Dispal® S250, which can be made by spray compacting only, has a melting

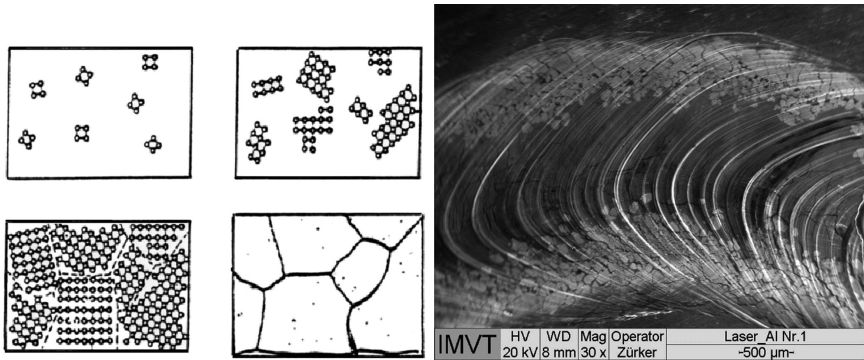


Figure 10.4 Left: Solidification of a melt (Pusch, 1983). Right: Cracks at the surface of AlMg3 after pulsed laser welding.

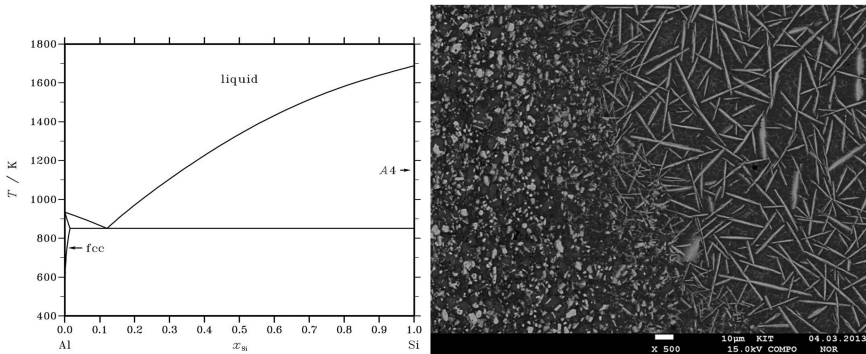


Figure 10.5 Left: Binary phase diagram of the system Al-Si (Europe, 2002). Right: Laser weld seam of AlSi20Fe5Ni2 (DISPAL® S250) by Gränges Powder Metallurgy GmbH, containing 20% Si. Laser welded at $P = 3$ kW, $v = 5$ m/min. Transition zone from needle-shaped primary Si-precipitates in the laser weld seam to spray compacted base material with globular precipitates.

interval of 572–780°C which is more than 200 K (Figure 10.5, left). AlMg3 (AW-5754), however, has a melting range from 610 to 640°C (30 K). Using the same parameters for laser welding as for S250, no crack-free weld seams could be obtained for AlMg3 (Gietzelt et al., 2014; Jia, Zhang, Yu, Shi, et al., 2021; Jia, Zhang, Yu, Wu et al., 2021; Zhang et al., 2021).

Laser Welding: Designs and Sequence

In micro process engineering, devices are often made of stacked thin sheet material containing mechanical microstructures. Especially for cw-laser weld

seams, bulging of the melt during solidification occurs. In this regard, in the end of the weld seam, a sink mark occurs.

For peripheral welding of multilayer stacks, shrinkage of the layers occurs at the rim in order to compensate bulging of the weld seam. Since the shrinkage is cumulative, it is not possible to weld all the layers in a strict consecutive way (compared to Figure 10.3). Otherwise, the residual gap is too large to be bridged. For this reason, the sequence and technology of welding become significant. Instead, for example, every tenth layer should be welded, and then welding should be continued staggered by a certain number of layers and/or changing the welding direction. This allows the heat dissipation to be more uniform, to minimize the thermal expansion, and to improve the accuracy of the weld seam position in terms of the pitch of the layers (Figure 10.3, right) (Gietzelt et al., 2013).

In order to weld thin sheets on the circumference with high process reliability, a sufficiently large joining cross-section is required. The connecting cross-section width should not be less than 0.4 mm at 50% of the welding penetration depth (Figure 10.6, left). This is to make sure for being able to compensate sheet thickness tolerances from cold rolling process, thermal expansion of the part during welding, as well as the positioning accuracy of the machine tool.

To weld thin sheet material or to integrate porous sintered metal layers into frames made of sheet metal, flat, wide weld seams are required. At the same time, a sufficient amount of melt must be generated to fill gaps (Figure 10.6, right) as it is explained in detail in Section 3.1 (Saha & Waghmare, 2020).

When welding through thin sheets, the sink mark at the end of a cw-weld seam where the plasma key hole collapses requires special attention. By appropriate design work, it should be placed outside a closed contour for guaranteeing vacuum gas tightness (Figure 10.7).

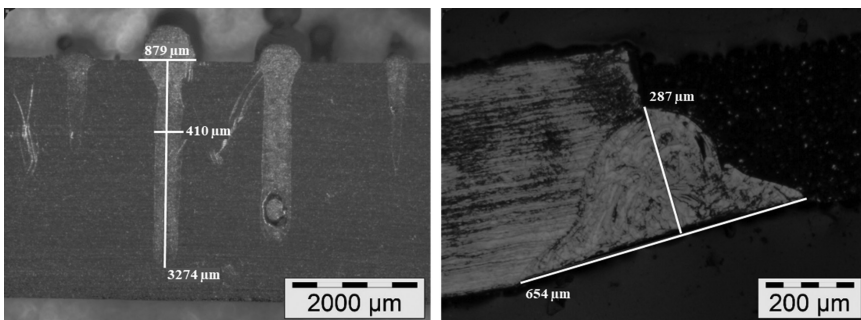


Figure 10.6 Left: Laser weld seam for circumferential welding of multilayer stacks.

$P_{\text{Peak}} = 2.5 \text{ kW}$, $P_{\text{av}} = 2 \text{ kW}$, $f = 333 \text{ Hz}$, $v = 4 \text{ m/min}$, focal position = -1 mm. Note bulging at surface. Right: Integration of porous sinter metal in sheet metal frames.

$P_{\text{Peak}} = 1 \text{ kW}$, $P_{\text{av}} = 400 \text{ W}$, $f = 400 \text{ Hz}$, $v = 2.4 \text{ m/min}$, focal position = +2 mm.

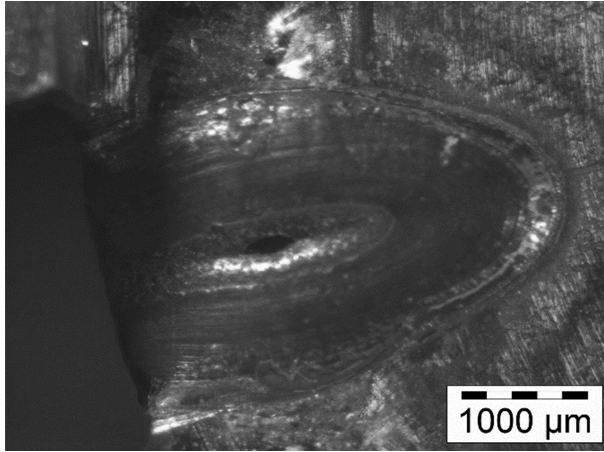


Figure 10.7 Sink mark and shrink hole at the end of a cw-laser weld.

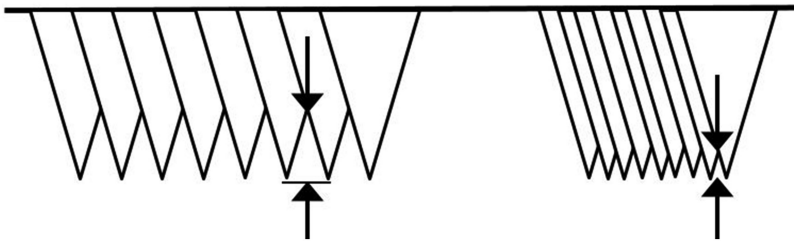


Figure 10.8 Relation of pulse overlap on welding penetration depth and vacuum tightness for pulsed laser welding.

Using pulsed laser welding, a sufficient overlap of single pulses must be guaranteed to achieve vacuum gas tightness (Figure 10.8). The required overlap of single pulses depends on the shape of the welding cross-section as well. Depending on the material welded, crack formation may arise since significant portions of material are remolten multiple times and subjected to rapid solidification (Chludzinski et al., 2021; Torabi & Kolahan, 2018).

By pulsed laser welding the energy input per unit of time or unit of length can be tailored very well by adapting average power, pulse length, and frequency to limit thermal distortion, especially for welding of thin sheet material (Gietzelt, Wunsch, Eichhorn, & Dittmeyer, 2015; Li et al., 2020; Xu et al., 2019).

Impact of Mechanical Microstructures

For micro process devices, mechanical microstructures have an impact on welding result. Depending on the thermal conductivity, the welding speed, and

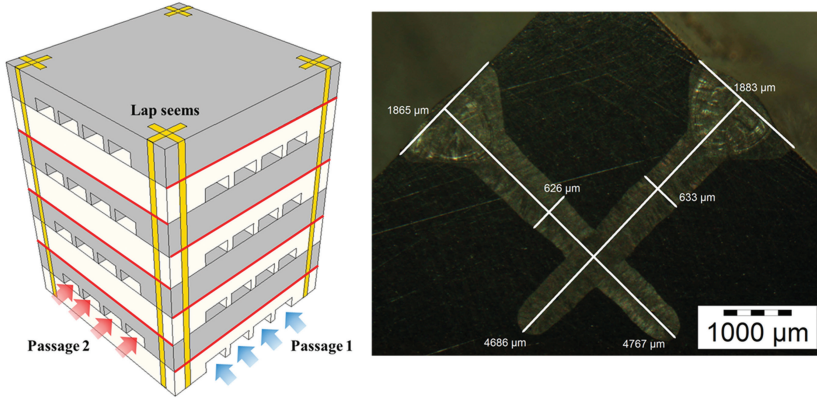


Figure 10.9 Left: Separation of different passages. Right: Cross-section of lap seams ($P = 3 \text{ kW}$, $v = 3 \text{ m/min}$, focal position = -2 mm).

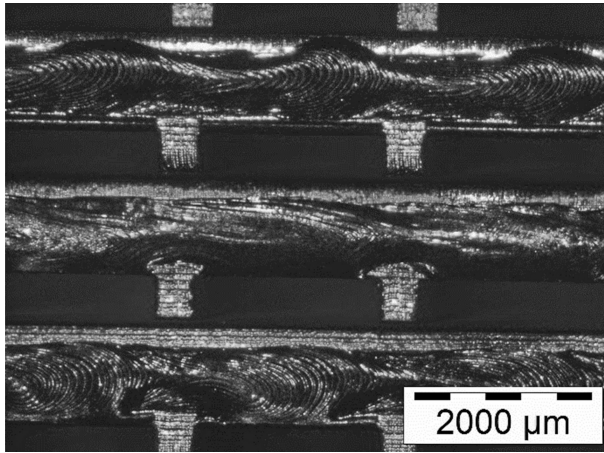


Figure 10.10 Impact of mechanical microstructures on weld seam geometry. Focal position = -1 mm , $P_{\text{peak}} = 1 \text{ kW}$. Top weld seam: $P_{\text{av}} = 300 \text{ W}$, $f = 400 \text{ Hz}$, $v = 2.4 \text{ m/min}$. Middle weld seam: $P_{\text{av}} = 750 \text{ W}$, $f = 1000 \text{ Hz}$, $v = 6 \text{ m/min}$.

the surrounding material into which the heat can be dissipated, spatial changes of the weld width can be observed (Figure 10.10).

The same applies to welding positions where the material accumulation differs considerably on both sides (Circular media feeds in Figure 10.15, top).

Laser Welded Micro Process Device for Hydrogen Separation

One of the most outstanding features of the laser welding which shed the light on the micro process devices or compact assemblies is the low local heat effects.

It becomes more vital in the chemical reactor engineering, e.g. when the active heterogeneous catalyst is already deposited into the reactor. In such a case, high temperatures reduce the activity of the catalyst or completely change its morphological properties. Another point of view is the integration of different layers for assembly and then jointing together. For instance, when it comes to integrating metallic foils, e.g. Pd-based membranes (few microns thick) with millimeter-thick supports/reactor, laser welding could be considered flexible and less aggressive compared to other techniques like diffusion bonding. In the following paragraphs, several cases are explained in detail.

Hybrid Metallic Supports for Composite Palladium-Based Membranes

Using Pd-based membranes is well studied and addressed in the last decades for hydrogen separation from a gas mixture. Although a Pd/Pd-alloy foil could be employed into a membrane without any support, the integration of a porous substrate with smooth surface has gained a lot of attention. Since non-metallic membrane supports come with challenges when joining and sealing, the metallic supports could be joined to the module/reactor by, e.g. welding. The design and gas sealing for such a structure requires a hybrid structure, e.g. a dense metal part (for integrating into other parts of the system) as well as that of a porous part (Figure 10.11).

It has to be emphasized that the joining technique requires a lot of attention not to damage the support structure, applicable for versatile alloys and conditions. For example, as planar metallic substrate, a tape-casted porous

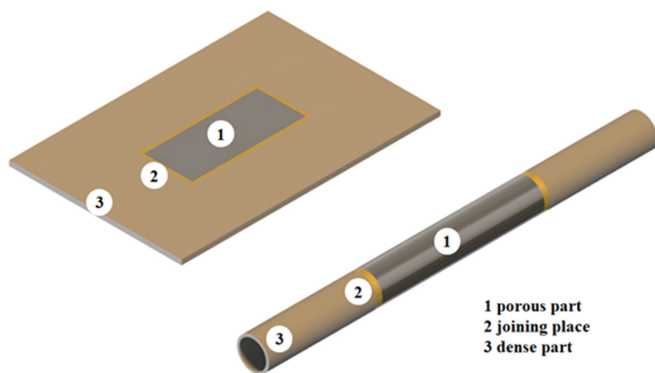


Figure 10.11 Joining metallic dense part to porous part for two common geometries using as membrane supports.

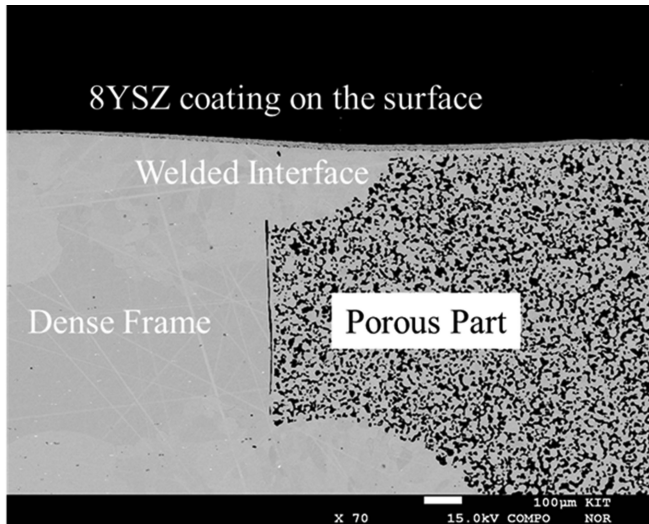


Figure 10.12 Laser welding for joining dense part to dense part for a metallic support membrane ($P = 800 \text{ W}$, $v = 3 \text{ m/min}$, focal position = 1 mm).

Crofer sheet is welded into a cold-rolled Crofer frame with laser welding. Figure 10.12 indicates a cross-section view of the (similar to plate structure in Figure 10.11) metallic membrane support (made of Crofer 22 APU, 1.4760) as an application for laser welding technique. The parameters for this welding are a power of 800 W, welding velocity of 3 m/min, and a focal position of +1 mm above the surface.

As shown in Figure 10.12, the technological challenge is to create an almost flat weld seam surface. Here, an equilibrium between the bulging of the weld seam and melting the porous part as well as filling the interface gaps must be achieved.

This is the prerequisite for being able to coat a thin and homogenous ceramic thermal barrier layer on top. It is to prevent diffusion of palladium into the metallic substrate at elevated temperatures during operation of the micro process device.

For this, a slurry of fully stabilized zirconia 8YSZ is used, which has a very close coefficient of thermal expansion to the ferritic stainless steel Crofer 22 APU (11 E-06/K) during subsequent sintering.

Integration of Thin Palladium-Membranes by Laser Welding

For gas separation application, thin membranes, e.g. consisting of palladium, can be integrated (Boeltken, Wunsch, Gietzelt, Pfeifer, & Dittmeyer, 2014).

For thin membranes, additional microsieve layers may be used as mechanical support. Therefore, several layers have to be welded together. Depending on the quality of the clamping, even the smallest air gaps between the layers provide effective thermal insulation (Figure 10.13, left). Consequently, the cross-section of the weld seam may be significantly reduced for bottom layers in comparison to welding results of same plate thickness without thermal insulation effects. Incorporating different material possessing different melting ranges and microstructures (nanocrystalline or amorphous) may affect molten areas locally (Figure 10.13, right).

Any void volume, e.g. due to use of grids or microsieves, may be filled by molten material. It may prevent bulging of the weld seam. and it, in turn, could save manufacturing of trenches for masking. However, the integrity of thin membranes is damaged by thermal insulation and surface tension effects (Figure 10.14).

Multilayer Reactor Coated with Catalyst

Reactors for heterogeneous catalysis could be fabricated in different shapes, e.g. tubular, and planar. It can be assembled with different techniques like using flanges, welding the parts, and so on. For compact structures consisting of different functioning plates as well as that with coated/deposited catalysts onto the reactor walls, laser welding technique might be beneficial. As mentioned earlier, as an outstanding feature, low local heat affects, when welding, prevents any deactivation of the deposited catalysts.

Figure 10.15 shows the proof of concept for this idea by which the number of the welded cells can be scaled-up. In order to verify the quality of the

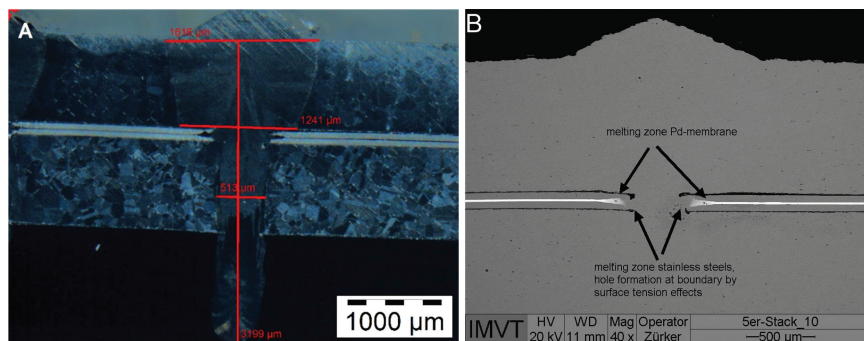


Figure 10.13 Left: Thermal insulation by air gaps between individual layers. Right: Extended melting range of the Pd membrane compared to the stainless steel layers.

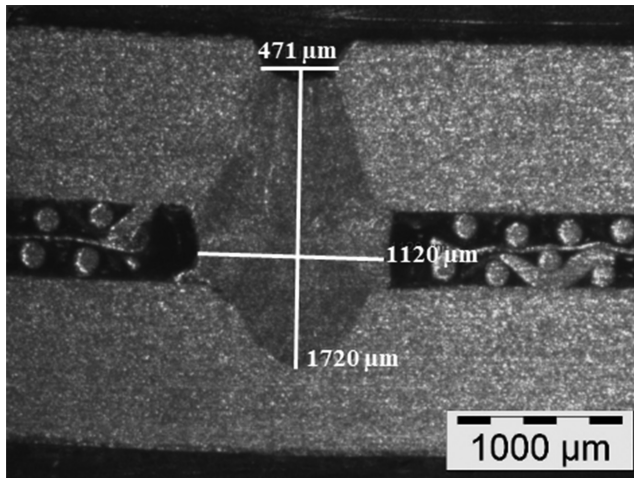


Figure 10.14 Integration of Pd-membrane 5 μm in thickness between stainless steel wire meshes (wire diameter 130 μm). $P = 2\text{ kW}$, $v = 4\text{ m/min}$, focal position = -1 mm.

welding, the welded parts are undergone a vacuum test (He leak test) and later a pressure test. In this study, TruLaser Cell 3010 (Trumpf, Germany) was used. The machine was capable of positioning accuracy of $\pm 15\text{ }\mu\text{m}$, a high dynamic of 10 m/s^2 of the axis, and a maximum speed of 50 m/min . It was equipped with a 3 kW Yb-YAG disk laser TruDisk 3001 (Trumpf, Germany), and a combined core-shell-lcc with $100/400\text{ }\mu\text{m}$ diameter. The focal length was $f = 150\text{ mm}$.

With such a multilayer design, it is necessary to have trenches, by which any bulging of the weld seams could be masked (yellow color in Figure 10.15). If the depth is identical to the depth of the microchannels, it can be carried out in one step, e.g. by chemical etching (as a cost-effective technique) without additional effort. For different depths, additional milling must be used, which is much more expensive.

In order to weld a multilayer stack, first two plates were welded from above, starting at the center plate. Then, the welding continued on the opposite side with the next plate, in order to prevent distortions from welding. To do so, a lot of design work is required to make sure trenches of subsequent layers must not cross each other.

By appropriate layout of the welding paths, the different gas passages could be separated from each other. As a critical step when assembly, the plate surface should be cleaned and clamped precisely to prevent any potential air gap. In order to verify the quality of the welding, the welded parts undergo a vacuum test (He leak test) and later a pressure test.

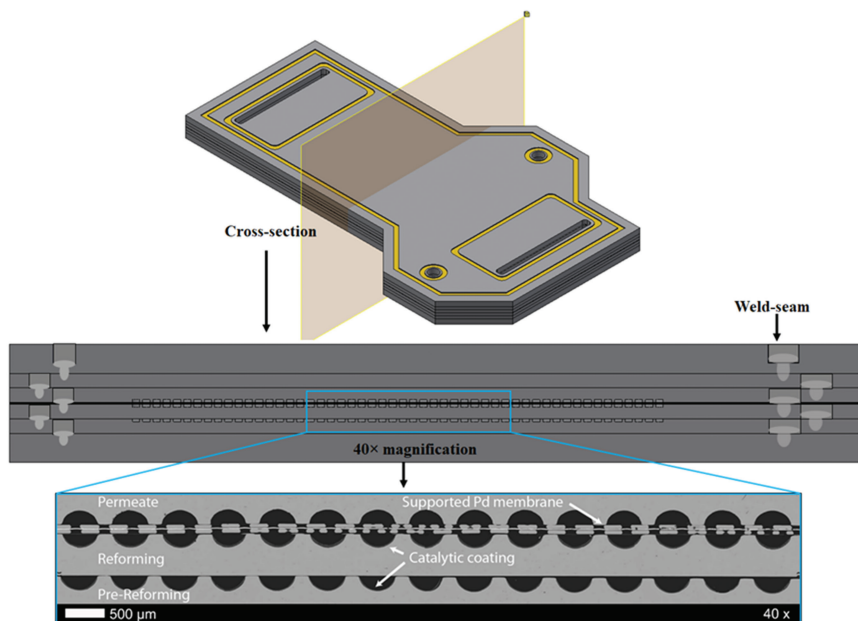


Figure 10.15 Multilayer ultra-compact membrane reformer (MicroEnH₂ancer V.1.0) for hydrogen-production and -separation joined by laser welding (Boeltken et al., 2014)[19].

Discussion

Due to the local and short duration of the heat input by laser welding, integration of thin parts consisting of different materials is possible. For diffusion bonding, however, diffusion processes take place, and grain growth occurs. For some materials, e.g. nickel-base alloys, the material microstructure is degraded in terms of corrosion properties through formation of precipitations at grain boundaries. This is due to the low thermal gradients in vacuum at high thermal masses, approaching thermodynamic equilibrium. Laser welding, however, enables a very high rate and short term of heat introduction, regardless whether pulsed or cw mode is used. Molten material solidifies very rapidly, far from thermodynamic equilibrium conditions.

In this study, it was shown that many additional factors for laser welding of micro process devices have to be considered, which usually can be ignored for macroscopic applications. Several preliminary tests are required to find out the unexpected effects. They are, for example, impact of mechanical microstructures on weld seam geometry and reduction of welding penetration depth for multilayer assemblies. Certain techniques should be implemented to meet

specific requirements, e.g. to join thin layers with mechanical microstructures or to successfully combine different materials. The cross-section of a laser weld seam in terms of width and welding penetration depth can be adapted in a wide range according to individual requirements. The smaller the lcc-diameter, the more sensitive is the laser welding process to technological variations like the focal position. If gaps between single sheet layers for circumferential welding arise, welding results for low lcc-diameters might vary strongly. This is in contradiction to that of larger lcc-diameters, where the percentage energy loss due to gaps is lower. Hence, the laser welding process becomes more stable for large lcc-diameters.

Conclusion

Since the costs per kilowatt laser power decreased considerably in the last decades, the integration of laser welding into process routes is easy and economic. Moreover, laser welding is widespread and well understood by many applicators, while only a few companies provide diffusion bonding as a service. Furthermore, the equipment for diffusion bonding is expensive, and the process time is rather long. Depending on the size of the parts to be bonded, costs can be divided, and multiple parts can be joined per run. It is worth mentioning that a certain deformation for diffusion bonding applies. It is difficult to control and it is not acceptable for all applications. Here, laser welding could offer several advantages. However, diffusion-bonded devices offer superior pressure resistance compared to laser welded parts. The design engineer must consider such constraints by picking the appropriate manufacturing and joining technologies to choose the most economic route for manufacturing of micro process devices.

The described special concerns are the reason why outsourcing the manufacturing of micro process devices to industry is difficult. At least, it must be accompanied individually by know-how transfer.

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Chapter 11

A Numerical Study on the Influence of Clamping on Laser Welding Residual Stresses and Distortions

F. A. O. Fernandes^{1,2}, A. B. Pereira^{1,2} and M.C. Silva¹

¹TEMA—Centre for Mechanical Technology and Automation, Department of Mechanical Engineering, University of Aveiro, Campus de Santiago, 3810-193 Aveiro, Portugal

²LASI—Intelligent Systems Associate Laboratory, Portugal

Abstract

Welding is a typical process employed to join a vast range of materials. This process originates residual stresses and distortions in the components, caused by the high energy input from the heat source. The use of numerical methods to simulate the welding processes has been increasing significantly. Numerical simulations make it possible to save time and are a more flexible methodology to predict and analyse complex phenomena. In this work, two numerical models were developed to predict residual stresses and deformations resulting from laser welding of Dual-Phase 600 steel plates in butt joint configuration. Different constraints were explored, determining the influence of clamping conditions on residual stresses and deformations.

Keywords: Laser welding, Numerical simulation, Welding distortion analysis, DP600, Residual stresses, Butt joint, Clamping

Introduction

Welding leads to distortions of the welded parts. The highly concentrated heat results in metallurgical changes, affecting mechanical behaviour. Laser

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welding is a fusion process characterised by a high and concentrated input energy resulting in many transformations (Kik 2020) and is usually performed in the absence of filler material.

Finite Element Analysis (FEA) makes it possible to predict the response of welded structures and improve the process by optimising process parameters. Nevertheless, the numerical simulation of laser welding is a complex process. There are several challenges to mimic the material transformations and also the accurate representation of the heat source.

FEA is a significant tool for uncovering the main challenges of the welding process, making it possible to determine the process output precisely. In other words, this makes it possible to minimise residual stresses and distortions in the weld, reducing costs and time (Caprace et al. 2017).

The simulation of welding processes usually can have many objectives but always aiming to improve the weld quality. For instance, to study the welding parameters' influence on the weld pool (Klingler 2018) or to quantify the residual stresses and distortions (Fernandes and Pereira 2021).

The knowledge of the material properties and how these change during welding can be employed to optimise the joining/process. One of the simulation essentials is to confirm the joint reliability without performing physical tests.

This chapter presents the development of two models. These were developed to simulate the laser welding process of dual-phase (DP) 600 steel plates. Both models focus on a butt joint configuration. These only differ in terms of constraints throughout the entire process, resulting in different boundary conditions. The resulting distortions and stresses were quantified.

Materials and Methods

The aim of this work is to simulate Nd:YAG laser welding of a butt joint with different clamping configurations. The resulting stress and distortion were analysed using the Simufact Welding.

Case Study – Butt Joint

This work aims to simulate the butt joining of two plates. This case study employs a Nd:YAG laser, with 4 kW and 3 m/min of power and scanning speed, respectively. The goal is to butt join two 20 mm square steel plates. These are 1 mm thick. As illustrated in Figure 11.1, there is contact between the sides to weld.

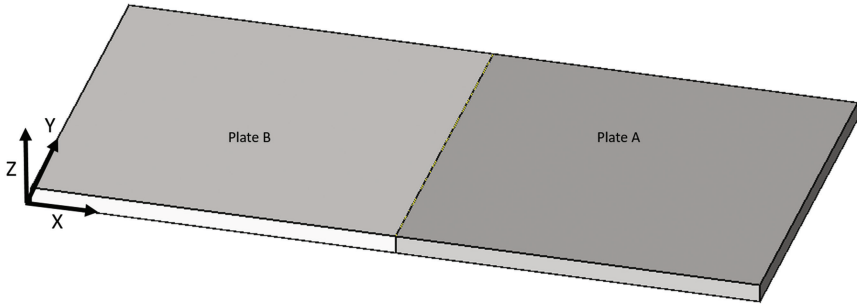


Figure 11.1 Butt joint configuration.

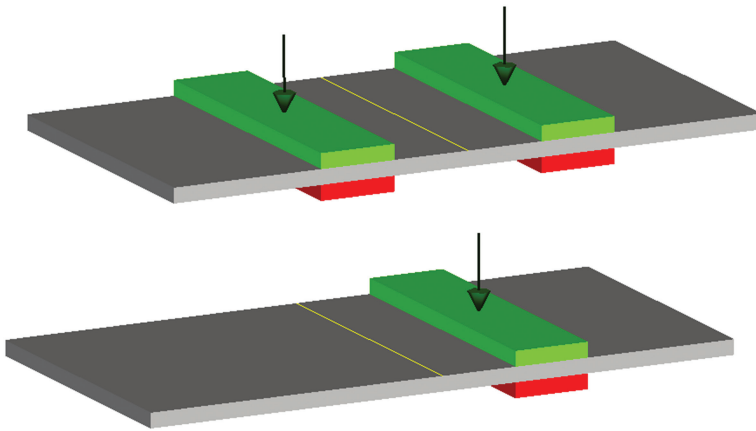


Figure 11.2 Model I – clamping conditions at step B (top) and steps A & C (bottom).

Setup of the Numerical Model

Each model has two steps: welding and cooling. The clamping system consists of rigid bodies with the following dimensions: $20 \text{ mm} \times 5 \text{ mm} \times 1 \text{ mm}$. Pairs of bearings-clampings were employed to constrain the plates during the welding. The bearing one (red in Figure 11.2) allows sliding on the same contact plane. The clamping system (green in Figure 11.2) presses the plates against the bearings with a 1 kN force and a constant translation stiffness of 5 kN/m.

The centre lines of these boundary conditions distance 7.5 mm from the bead's centre. Table 11.1 summarises the changes in the boundary condition during the entire simulation, resulting in two models. Model I is developed to determine the maximum distortions, while model II is designed to determine the maximum stress.

Room temperature was set to 20°C , and gravity was considered (9.80665 m/s^2). The cooling step is activated when the welding is over and

Table 11.1 Models I and II clamping conditions

Model	Aim	Step		Plate A		Plate B	
				Clamp	Bearing	Clamp	Bearing
I	Max. deformation	B	Welding (tack)	ON	ON	ON	ON
		C	Welding (joining)	ON	ON	OFF	OFF
		A	Cooling	ON	ON	OFF	OFF
II	Max. stress	B	Welding (joining)	ON	ON	ON	ON
		C	Cooling	ON	ON	ON	ON

Table 11.2 Incrementation between analysis

Parameter	Welding step	Cooling step
Mechanical solution every n thermal increment	3	5
Result output every n increment	6	5

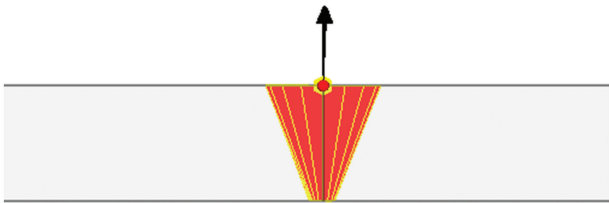


Figure 11.3 Heat source shape.

only ends when the workpiece decreases to 300 °C. This simplification has a negligible effect on the results and minimises the computational time necessary that would take to achieve room temperature.

The analysis was performed in precision mode, and both thermal and mechanical analyses were performed together. Phase transformations were disregarded. The total processing time for all models was 35 seconds to ensure the cooldown phase after the end of the welding phase. Table 11.2 presents the values employed to define the frequency of mechanical calculation relative to thermal analysis and the results output frequency for welding and cooling cycles.

Heat Source

Figure 11.3 depicts the conical heat source with Gaussian distribution in key-hole mode employed to simulate the laser welding. Table 11.3 presents the dimensions of the heat source, characterised by uniform power distribution.

Table 11.3 Heat source dimensions

Parameter	Value [mm]
Radius superior (r_u)	0.5
Radius inferior (r_l)	0.1
Depth (d)	1.0
Offset	0

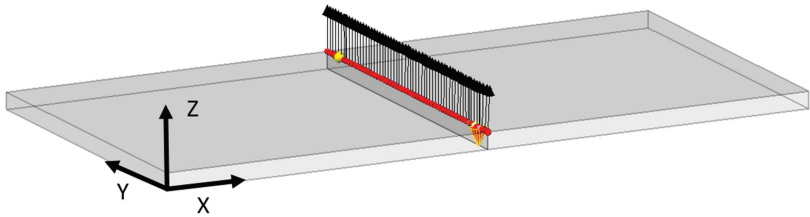


Figure 11.4 Trajectory of the heat source.

Figure 11.4 presents the trajectory of the heat source defined to perform the butt joint. The weld bead has a total length of 20 mm. Tack welds have a length of 1 mm.

Material

The material properties available in the material library of Simufact Welding (S60QL-SPM-sw) and with similar properties to DP600 steel were employed. The initial phase fractions assume 79% bainite and 21% ferrite. The material presents a solidus and melting temperatures of 1462.92 °C and 1509.95 °C, respectively. The latent heat for melting has a value of 256.4 kJ/kg. The thermal conductivity is temperature-dependent, as described in Figure 11.5. Figure 11.6 depicts the material's specific heat capacity thermal dependency.

Regarding the mechanical properties, the Poisson's ratio was defined with a constant value of 0.3. The Young's modulus temperature dependency is depicted in Figure 11.7.

Figure 11.8 depicts the density dependency on the temperature, while Figure 11.9 shows the temperature dependency of the thermal expansion coefficient, which describes the material's volumetric changes with temperature.

Figure 11.10 shows the flow curves at different temperatures (25 °C, 100 °C, 200 °C, 300 °C, 400 °C, 500 °C, 600 °C, 700 °C, 800 °C, 900 °C, 1000 °C, 1100 °C, 1200 °C and 1300 °C), necessary to define the true stress – true plastic strain relationship for several temperatures. As the temperature increases, the lower the stress level is.

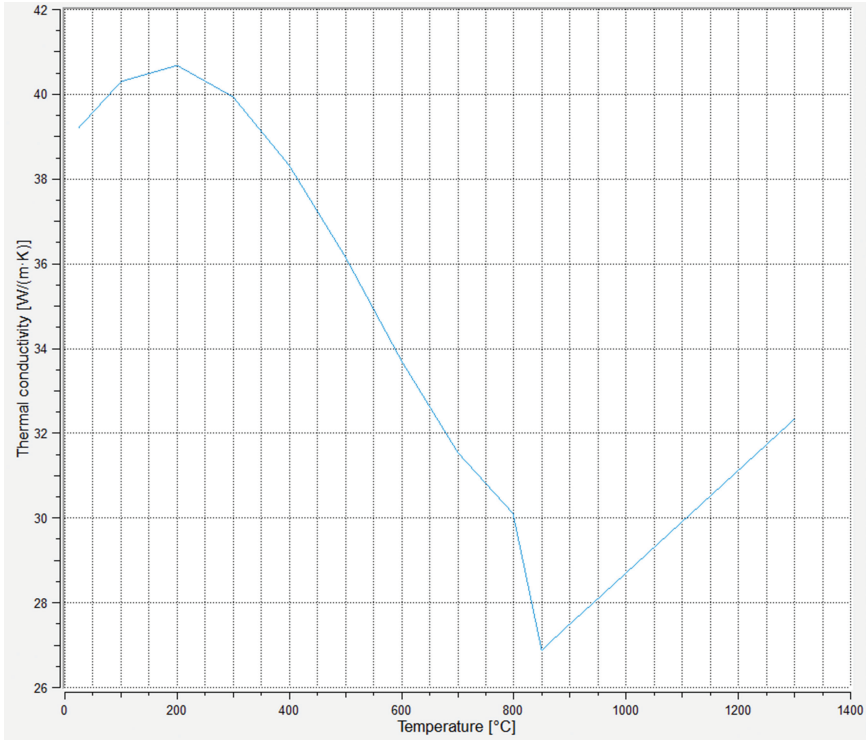


Figure 11.5 Thermal conductivity vs temperature.

Mesh

A mesh convergence study was conducted to mitigate the mesh influence on the results and to avoid excessive mesh refinement, which would result in excessive computational time. The final model is sectioned into three regions, from less to more refined mesh: base material, heat affected zone and fusion zone. The final model is composed of 11544 hexahedral finite elements with 3 elements across the thickness, and 8160 nodes (Figure 11.11). The elements' size ranges from $0.25 \text{ mm} \times 0.25 \text{ mm} \times 0.33 \text{ mm}$ to $1.0 \text{ mm} \times 1.0 \text{ mm} \times 0.33 \text{ mm}$.

Results

Residual Stresses

Two output lines have been defined to quantify the stresses in the transverse (x) and longitudinal (y) directions. The 20 mm longitudinal line matches the

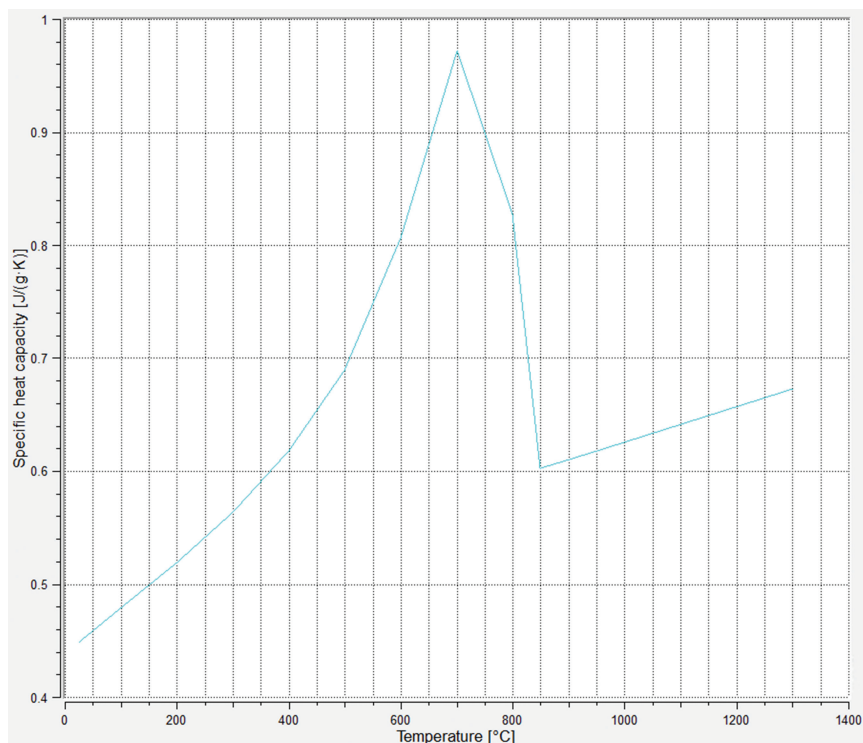


Figure 11.6 Specific heat capacity vs temperature.

bead. The 40 mm transverse line is perpendicular to the bead and positioned in the plate's middle. Figure 11.12 illustrates the lines positioning. There are 40 tracking points for each mm.

As illustrated in Figure 11.13, the lines are positioned on both top (orange) and bottom (green) surfaces in order to compare the stress across the thickness. In total, each model has four lines to monitor the stress distribution.

Displacement Measurement

Three transverse lines (a, b and c, following the order of the welding direction, and b being positioned in the middle) and two longitudinal lines (lines d and e) were set (Figures 11.14 and 11.15). The transverse and the longitudinal lines have a length of 40 mm and 20 mm, respectively.

The lines set to determine the displacements are only placed on the top surfaces. Since the employed plates are 1 mm thick, it is reasonable to consider

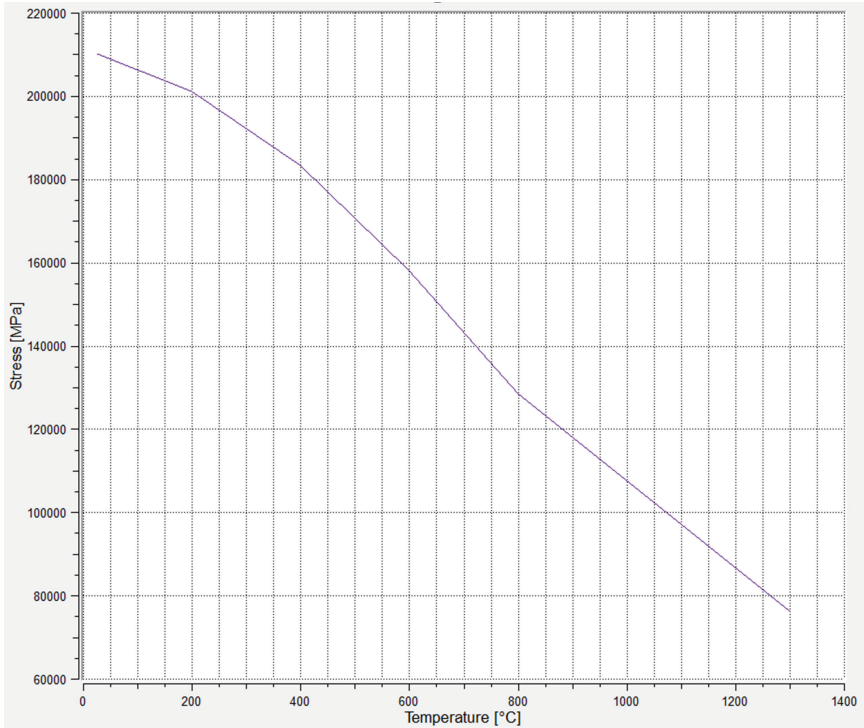


Figure 11.7 Young's modulus vs temperature.

the same deformation on both surfaces. Table 11.4 summarises the output and location of measurement for both models.

Results – Model I

Longitudinal Stress

Figure 11.16 shows an isometric view of the longitudinal stress distribution. The weld bead area has a higher stress concentration because heat causes material expansion, originating residual stresses after cooling. The longitudinal stress is identical on both top and bottom surfaces, and its distribution is symmetrical.

Figure 11.17 depicts the longitudinal residual stress along the transverse lines. Minimum values are measured on the edges since these are just slightly affected by the heat during welding. The plates are under a compression state in the middle, approximately around -225 MPa. Near the bead (± 5 mm from the weld line), it is under a high tensile state.

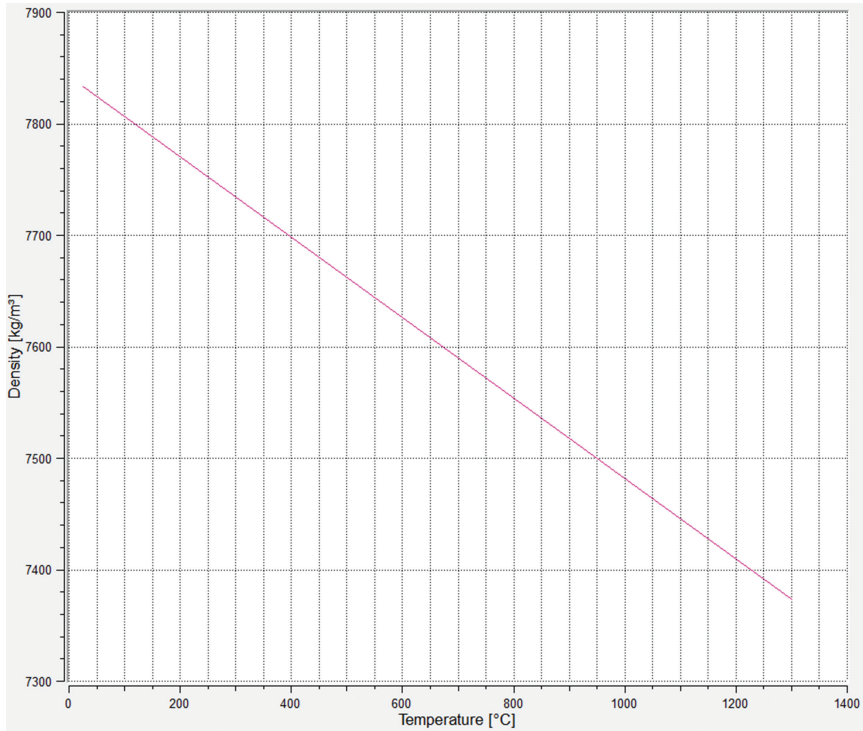


Figure 11.8 Density vs temperature.

The peak close to 730 MPa is in the middle of the plates, which corresponds to the melted zone. This happens due to the higher temperatures, resulting in significant microstructure transformations. Also, since all the clamping conditions are active during both welding and cooling steps, it significantly influences the magnitude of the residual stresses.

Figure 11.18 reveals the presence of longitudinal stresses along the bead. The results show a small disparity between both surfaces regarding stress magnitude and distribution along the line. The fact that the top surface is exposed to higher heat due to the direct interaction with the laser beam is a potential justification for these differences. On the edges, null stress is measured. Nevertheless, higher stresses are reached closer to the bead, up to a maximum value of 710 MPa.

Transverse Stress

Figure 11.19 shows that the clamping conditions greatly influence transverse stress. The highest stress concentration is located within the bead and the clampings. Beyond these regions, there is a significantly decreased stress since

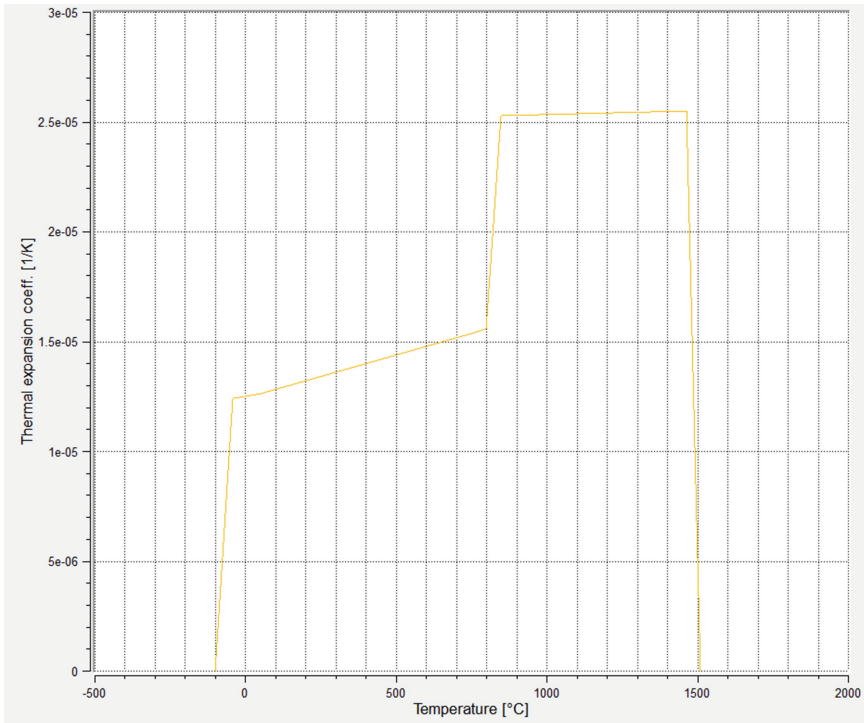


Figure 11.9 Thermal expansion coefficient vs temperature.

these are allowed to expand during welding and shrink during the cooling phase.

Transverse stress is practically the same between both surfaces, achieving a slightly lower magnitude at the bottom one, which is expected since the opposite surface absorbs the laser. The profile of the transverse residual stress is symmetrical. Figure 11.20 shows maximum stress along the transversal line in the bead region, reaching a magnitude around 650 MPa. A tensile state (+) is present along the transverse direction. Also, there is a small state of compression (-) in a small area near the end of the clampings, reaching null transverse stress at the edges.

Figure 11.21 highlights a compression state at the edges of around -450 MPa and -600 MPa. Outside, the bottom surface has a tensile state of over 600 MPa and almost 450 MPa of maximum transverse stress on the top.

Out-of-Plane Deformation (z-direction)

Figures 11.22 and 11.23 show the displacement in the z-direction, null in the clamping regions. Figure 11.24 shows the deformation in the z-direction, which is symmetrical (joint symmetry plane). As expected, the deformation is null in

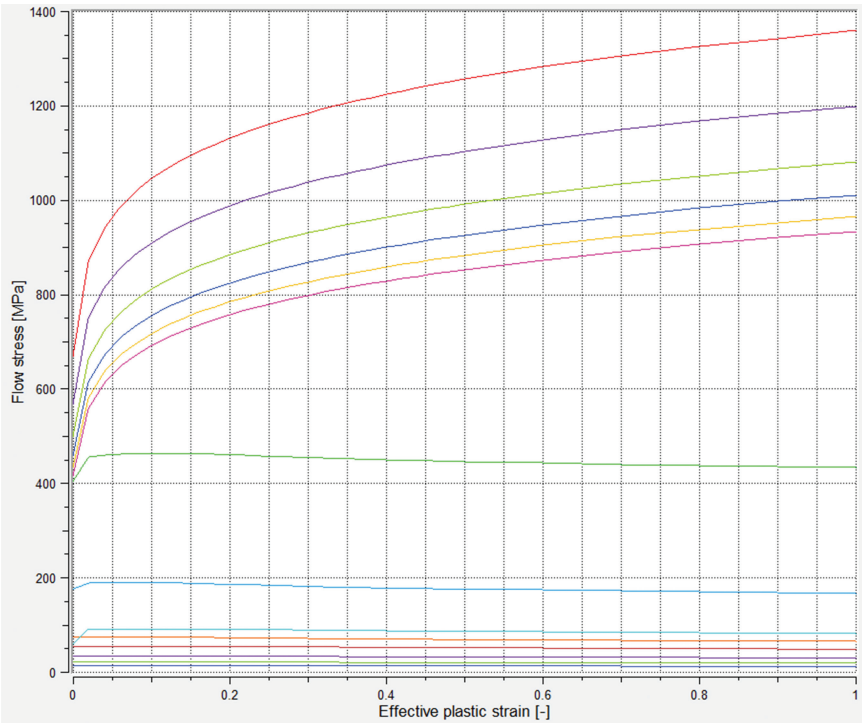


Figure 11.10 Flow curves (the stress levels decrease with temperature evolution).

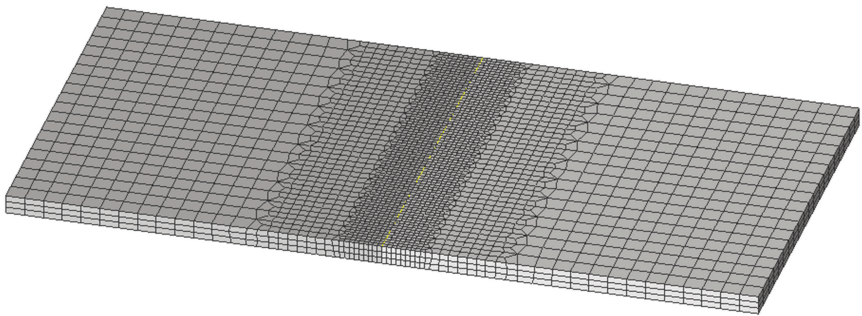


Figure 11.11 Final mesh obtained after the convergence study.

the constrained region. The deformation at the edges is around 0.05 mm. The deformation at the beginning and end of the bead is similar, with a displacement of around -0.02 mm.

Longitudinal Deformation (y-direction)

Figure 11.25 presents a top view of the sample, showing the results for displacement in the y-direction. Excluding the bead centre, the edges of the plate

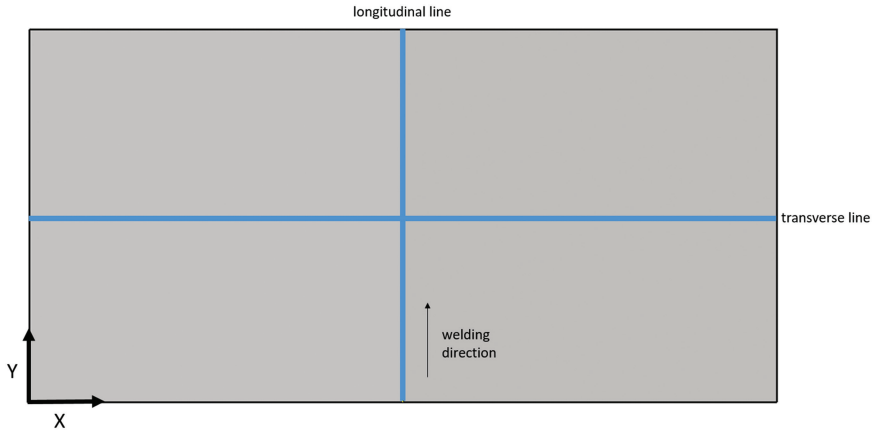


Figure 11.12 Top view of the models - transverse and longitudinal lines positions.

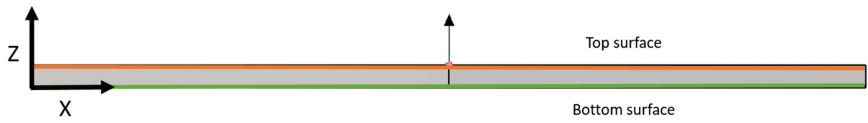


Figure 11.13 Top and bottom surface lines.

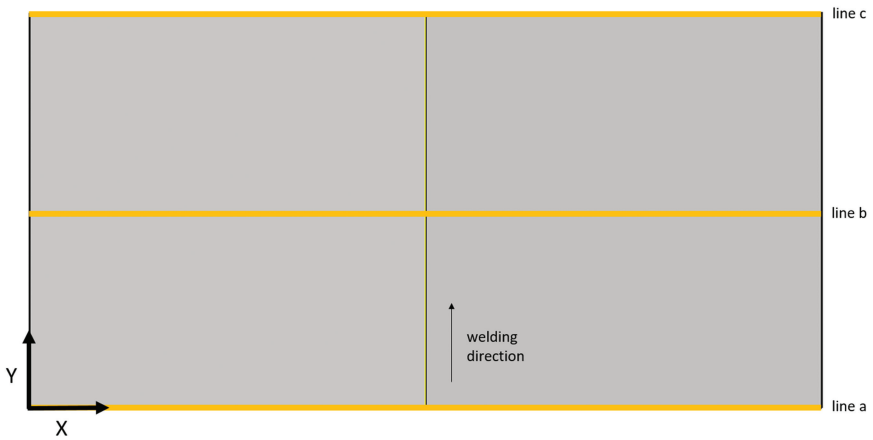


Figure 11.14 Transverse lines to determine Y- and Z-displacements.

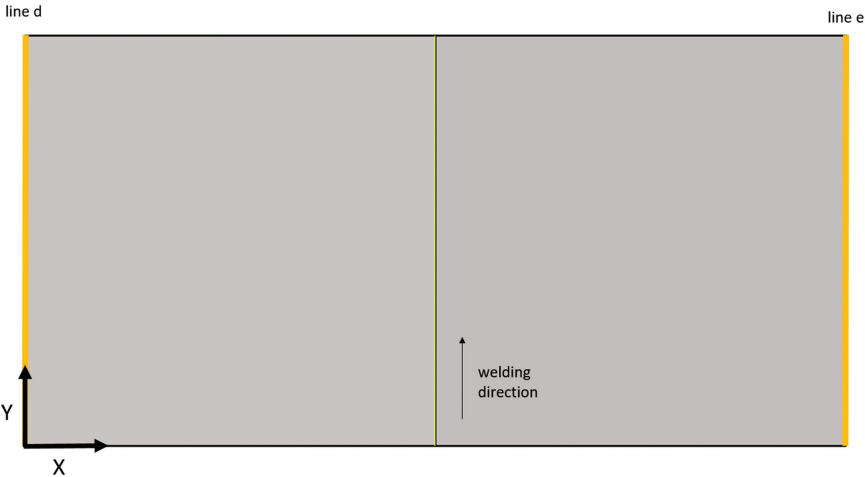


Figure 11.15 Longitudinal lines to determine the X-displacement.

Table 11.4 Summary of the results for both models

Parameter	Location
$\sigma_{\text{longitudinal} - xx}$	Transverse (both surfaces)
	Longitudinal (both surfaces)
$\sigma_{\text{transverse} - yy}$	Transverse (both surfaces)
	Longitudinal (both surfaces)
Out-of-plane deformation (ϵ_{zz})	Transverse (lines a, b, c)
Longitudinal deformation (ϵ_{yy})	Transverse (lines a, b, c)
Transverse deformation (ϵ_{xx})	Longitudinal (lines d, e)

in the y-direction (lines a and c) have a symmetrical longitudinal shrinkage with a magnitude of 0.04 mm, as depicted in Figure 11.26.

Transverse Deformation (x-direction)

Figure 11.27 shows the deformation in the x-direction. The largest deformation occurs between the bead and the clamps. The deformation that exists in the x-direction is symmetrical. Outside this zone, the deformation is insignificant. This deformation directly influences the plate length. Nevertheless, the length change is minimal, increasing less than 0.02 mm, as depicted in Figure 11.28.

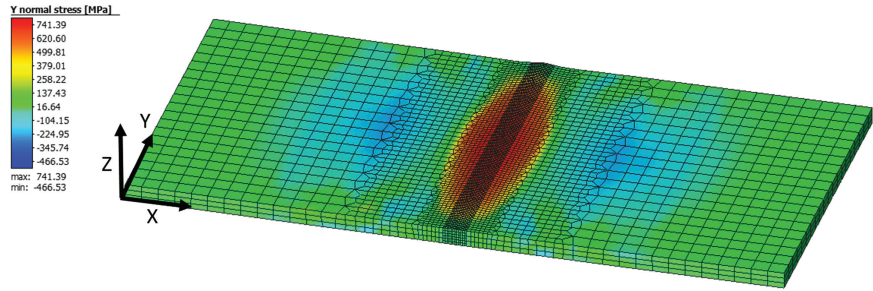


Figure 11.16 Longitudinal stress mapping.

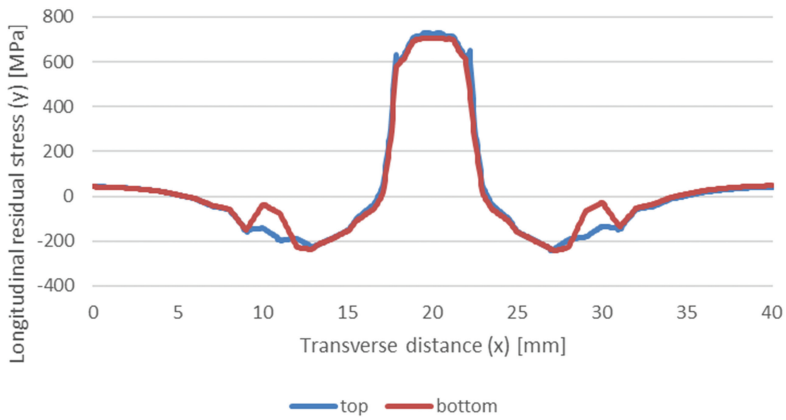


Figure 11.17 Longitudinal residual stress along transverse lines.

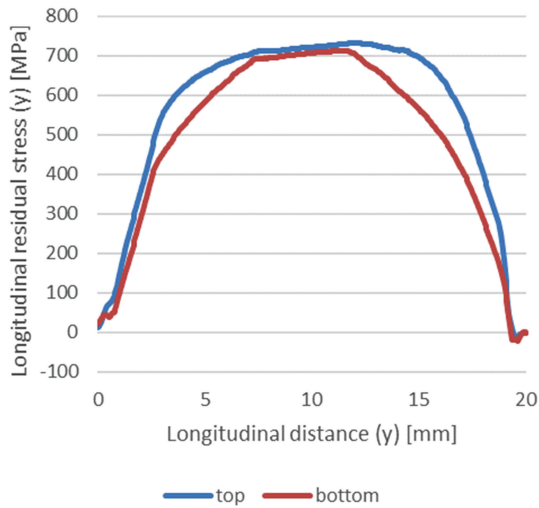


Figure 11.18 Longitudinal residual stress along longitudinal lines.

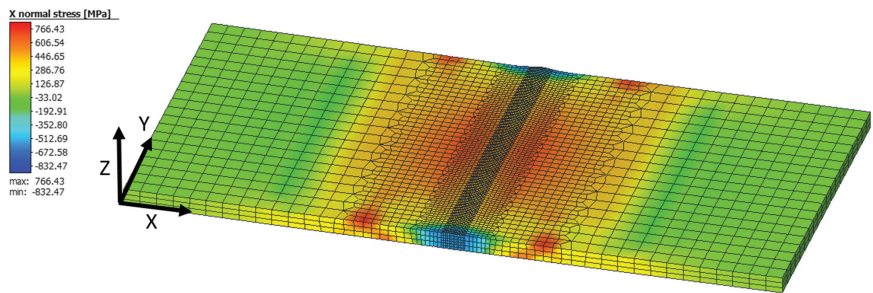


Figure 11.19 Transverse stress mapping.

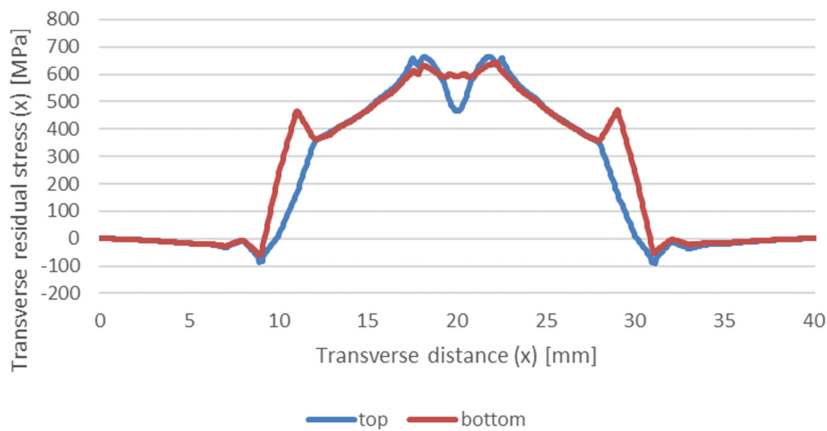


Figure 11.20 Transverse stress along transverse lines.

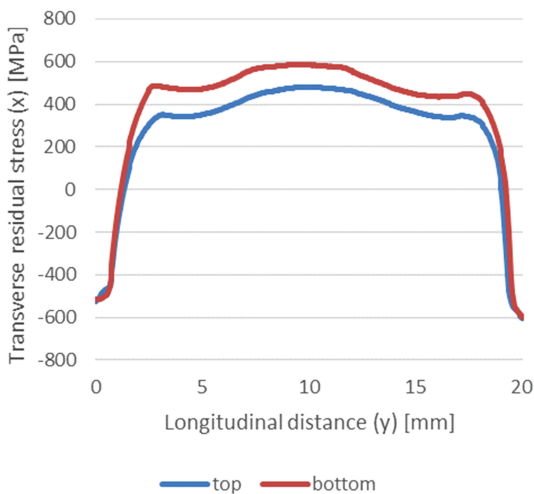


Figure 11.21 Transverse stress along longitudinal lines.

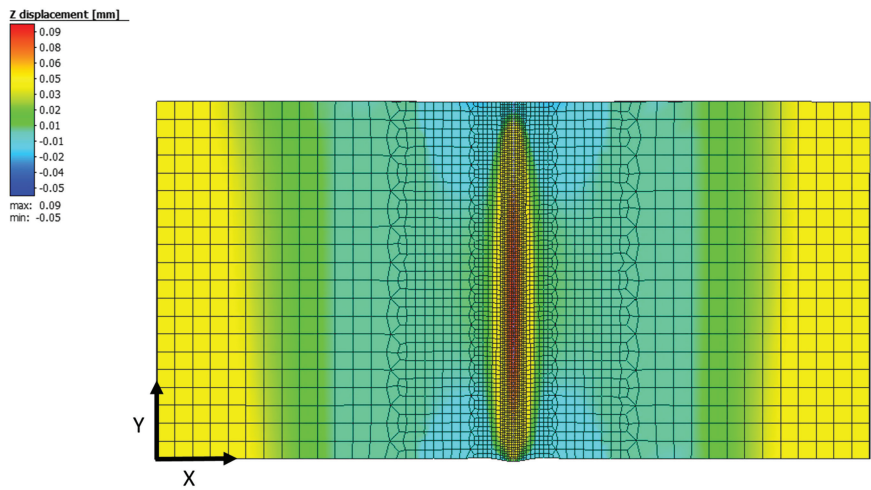


Figure 11.22 Top view – Z-displacement.

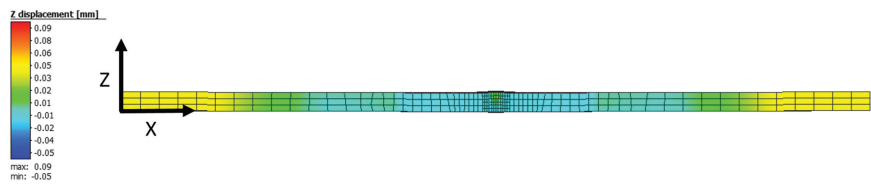


Figure 11.23 Side view – Z-displacement.

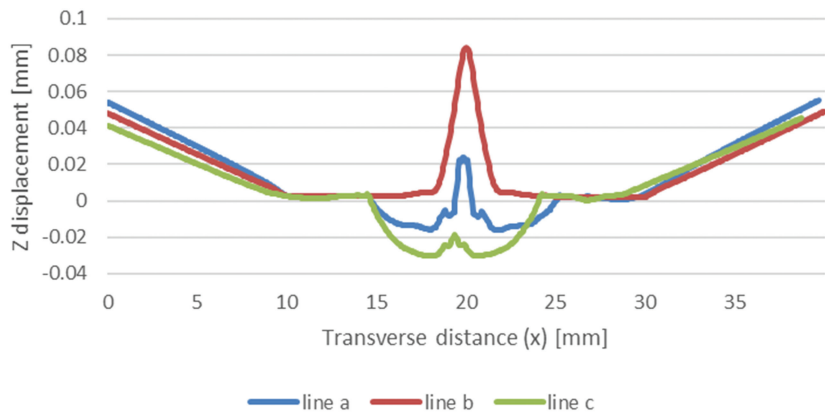


Figure 11.24 Z-displacement along transverse lines.

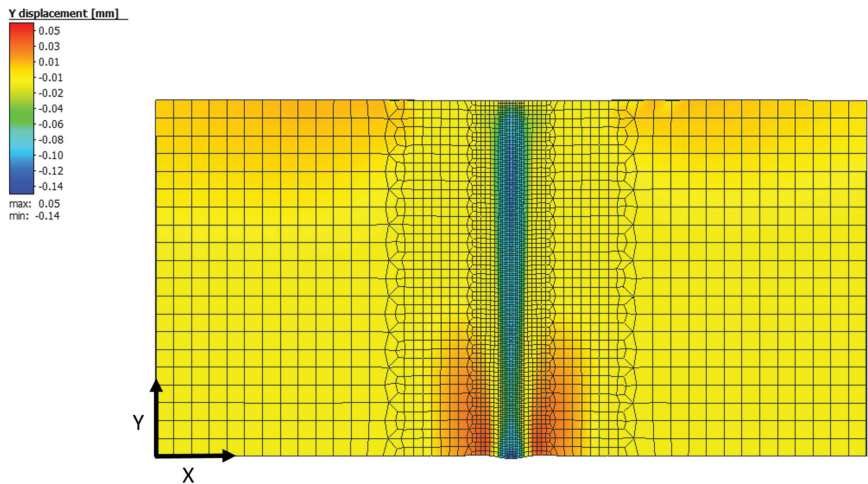


Figure 11.25 Top view – Y-displacement.

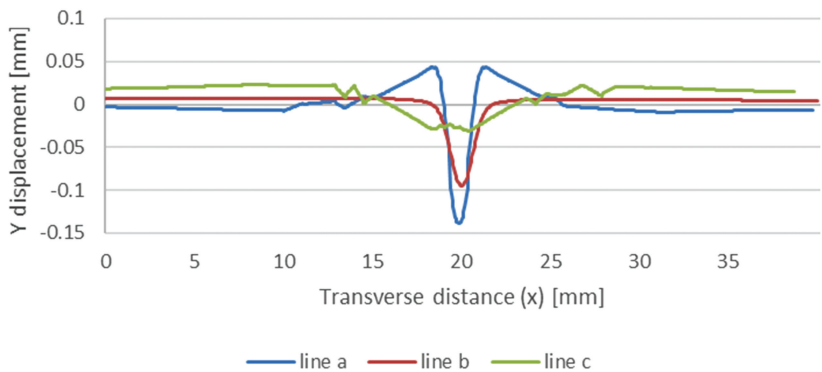


Figure 11.26 Y displacement along transverse lines.

Results – Model II

Longitudinal Stress

Figure 11.29 depicts higher longitudinal stress near the weld bead. Figure 11.30 shows a similar magnitude on the top and bottom surfaces and also symmetry. Around the bead, there is a tensile state, while outside this region, it is under compression. The maximum and minimum residual stresses in the transverse direction are around 800 MPa and -450 MPa, respectively.

The longitudinal stress values are identical on both top and bottom surfaces, with a slight difference between top and bottom surfaces at the edges.

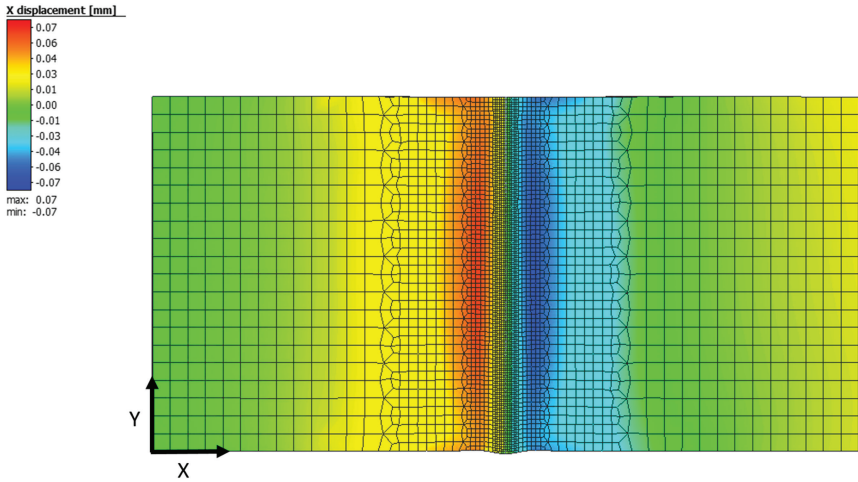


Figure 11.27 Top view – X-displacement.

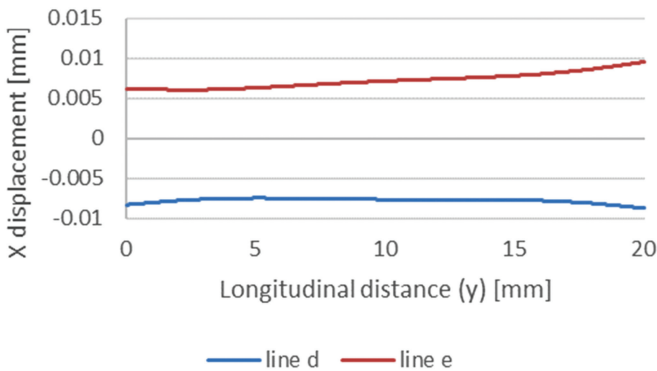


Figure 11.28 X-displacement along longitudinal lines.

Figure 11.31 depicts a symmetric residual stress profile, with a maximum value of 710 MPa at the centre of the plate.

Transverse Stress

Figure 11.32 shows the predicted residual stresses of the welded component. There are some noticeable differences since plate B ($0 \leq x \leq 20$ mm) is not constrained during welding. The transverse stress along transverse lines differs between both surfaces, and no symmetry is observed between plates. This is evident in Figure 11.33, where Plate A ($20 < x \leq 40$ mm) presents more stress variability, probably because it is constrained during the entire process. Plate

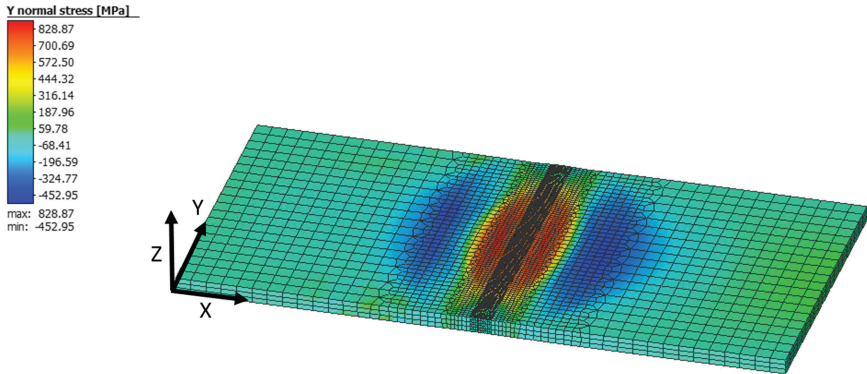


Figure 11.29 Mapping of longitudinal stress.

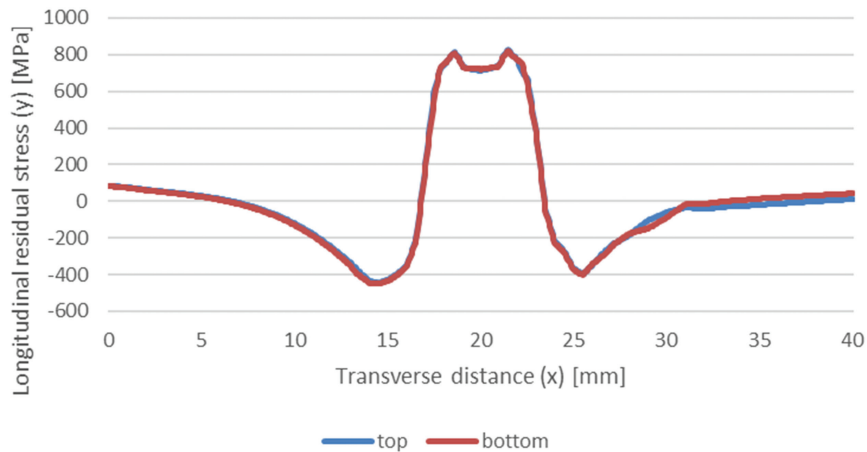


Figure 11.30 Longitudinal residual stress along the transverse lines.

B presents a state of compression with a peak value of around -50 MPa. Closer to the bead, the residual stress magnitude in the transverse direction increases up to a peak value of approximately 250 MPa. Plate B is characterised by a more significant compression state and with a higher disparity between the top and bottom surfaces, reaching a minimum value of around -210 MPa for the latter.

The transverse stress profile along the longitudinal lines is symmetrical at the bead and with minimum differences between surfaces. In the extremities of the bead, there is a compression state with peak values of approximately -650 MPa. On the other hand, as depicted in Figure 11.34, the maximum positive value of approximately 200 MPa is achieved in the middle of the bead.

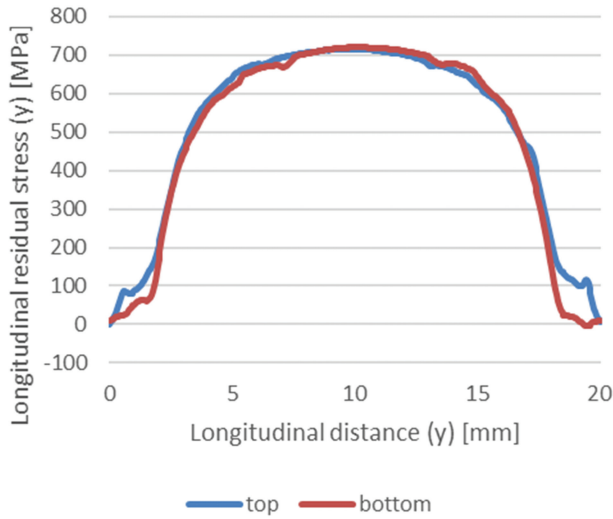


Figure 11.31 Longitudinal residual stress along the longitudinal lines.

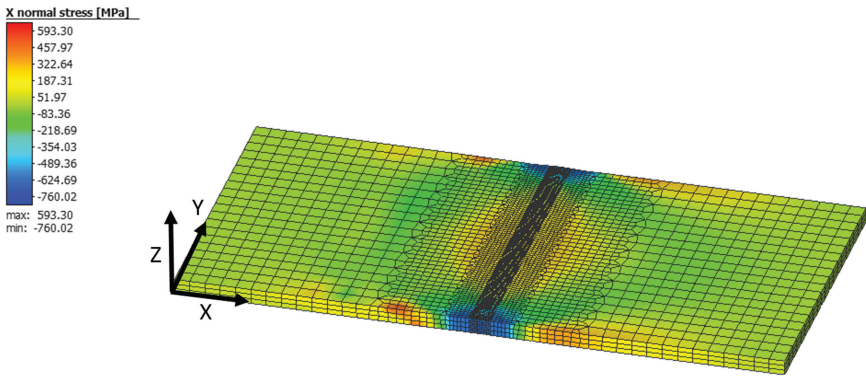


Figure 11.32 Mapping of transverse stress.

Although the transverse tensile stress has a lower magnitude than the compression one, most of both plates are under a tension state.

Out-of-Plane Deformation (z-direction)

Figures 11.35 and 11.36 show that the out-of-plane deformation is minimal. Although not clamped during welding, plate B is characterised by low out-of-plane. The largest out-of-plane deformation is predicted to occur in the bead.

Figure 11.37 presents the deformation along lines a, b and c. Plate B is characterised by negative and null displacement values closer to the bead and

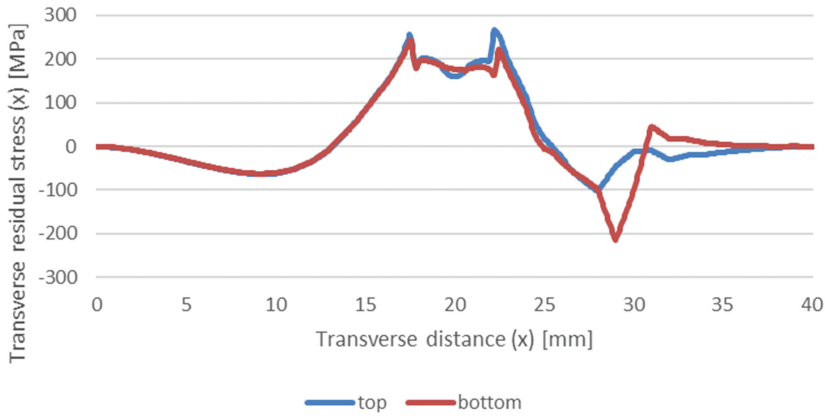


Figure 11.33 Transverse residual stress along the transverse lines.

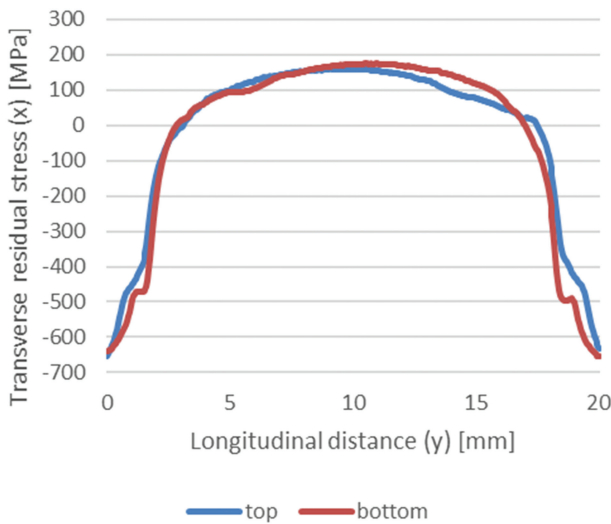


Figure 11.34 Transverse residual stress along the longitudinal lines.

at its beginning, respectively. Plate A presents null out-of-plane displacements in the clamped regions and -0.01 mm outside this region.

Longitudinal Deformation (y-direction)

Figure 11.38 reveals rotation distortions, probably caused by the absence of clamping on plate B. Thus, the transverse shrinkage is not uniform along the bead. There is a more significant shrinkage at the bead's end than at its beginning, meaning a high welding speed.

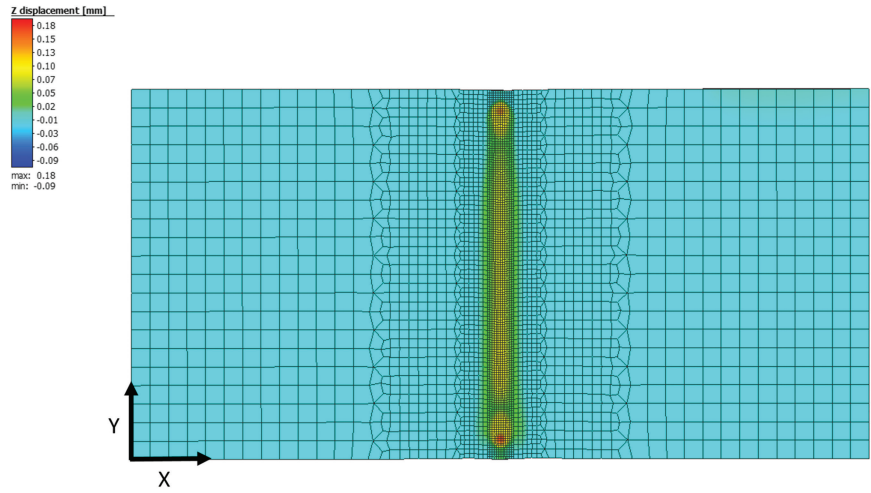


Figure 11.35 Top view – Z-displacement.

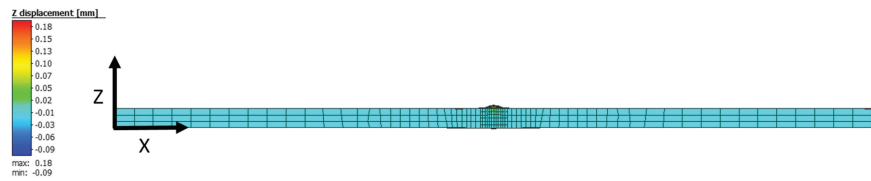


Figure 11.36 Side view – Z-displacement.

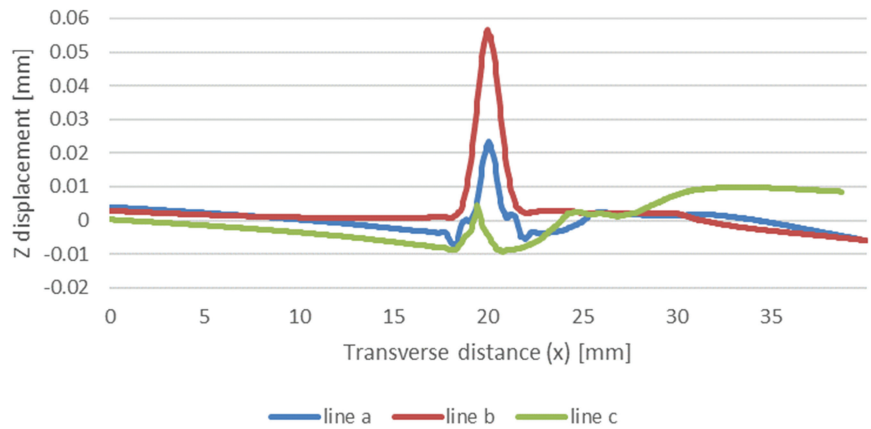


Figure 11.37 Z-displacement along transverse lines.

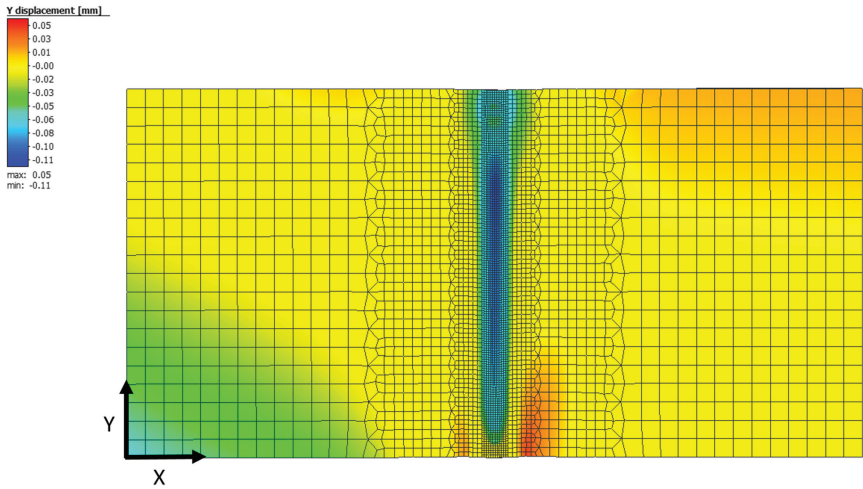


Figure 11.38 Top view – Y-displacement.

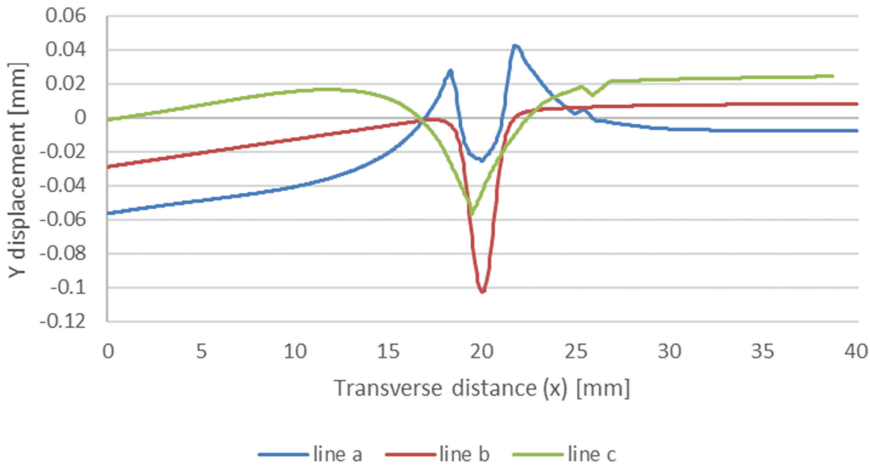


Figure 11.39 Y-displacement along transverse lines.

The longitudinal deformation is significantly higher on plate B. Figure 11.39 reveals that plate B has a peak displacement of almost -0.06 mm. The behaviour along the width is not uniform. The welding direction has an evident influence on the predicted longitudinal deformation due to the evolution of the values along the bead. Minimal displacements are achieved in the clamped region of plate A. Nevertheless, the most thermally affected zone has a symmetrical longitudinal shrinkage.

Transverse Deformation (x-direction)

Transverse deformation influences the final length of the welded component. As expected, due to the model’s boundary conditions, the transverse deformation is not symmetrical between plates (Figure 11.40). Figure 11.41 depicts the x-displacements. Along line e, there is an almost uniform displacement of around 0.017 mm. Also, the results show a rotational distortion of plate B.

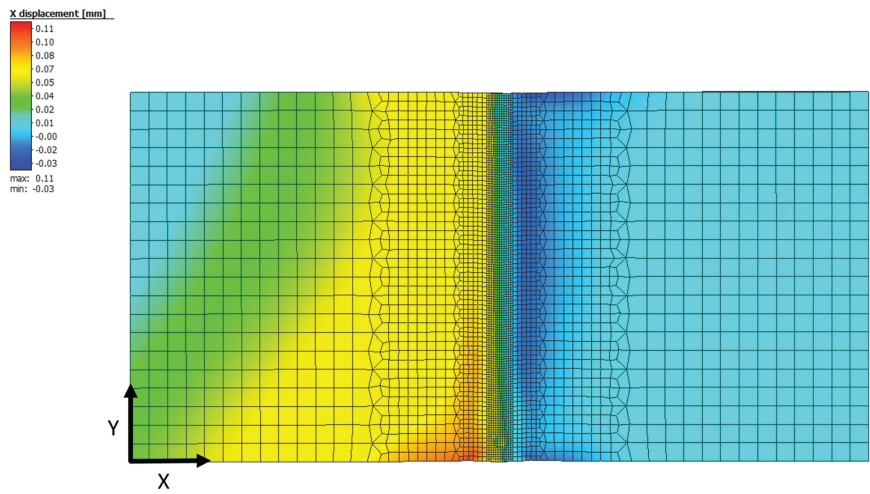


Figure 11.40 Top view – X-displacement.

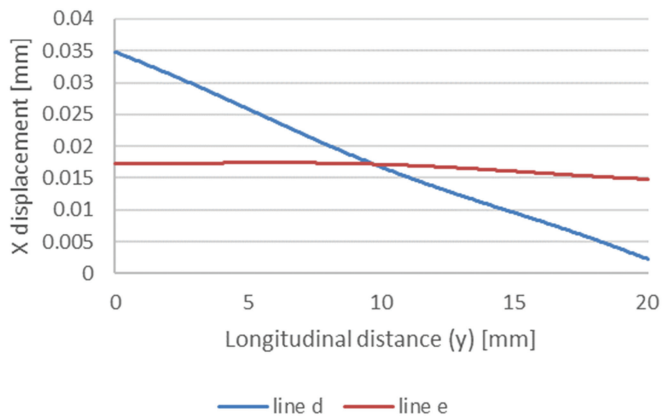


Figure 11.41 X-displacement along longitudinal lines.

Discussion

Figure 11.42 shows that the longitudinal stress is similar in both models. Thus, the influence of the present clamping conditions on the longitudinal stress is small. In both cases, compressive and tensile states are present in the regions that are less and more affected by heat, respectively. In addition, the higher longitudinal stress is measured next to the bead. This is also justified by the higher temperature gradient in this region. Also, the differences between measurements on both top and bottom surfaces are negligible.

The longitudinal stress along the longitudinal lines has a higher magnitude than the one measured along the transverse lines. Based on Figure 11.43, it is possible to confirm that these are in the tensile state, reaching the maximum at the bead's centre. Again, the superior temperature gradient is behind the greater residual stresses.

Regarding the transverse stress, there is a significant discrepancy between both models' predictions, meaning it is dependent on the clamping conditions. The full constrained model achieved higher transverse stresses.

From Figure 11.44, it is possible to conclude that transverse stresses are also more affected by the temperature gradient ($10 < x < 30$ mm), independently of the model. In the region of the weld bead, there is a maximum difference of approximately 400 MPa between model I and model II. It is also worth highlighting that in model II, there is a compression state (-) between joint and the plate's extremity.

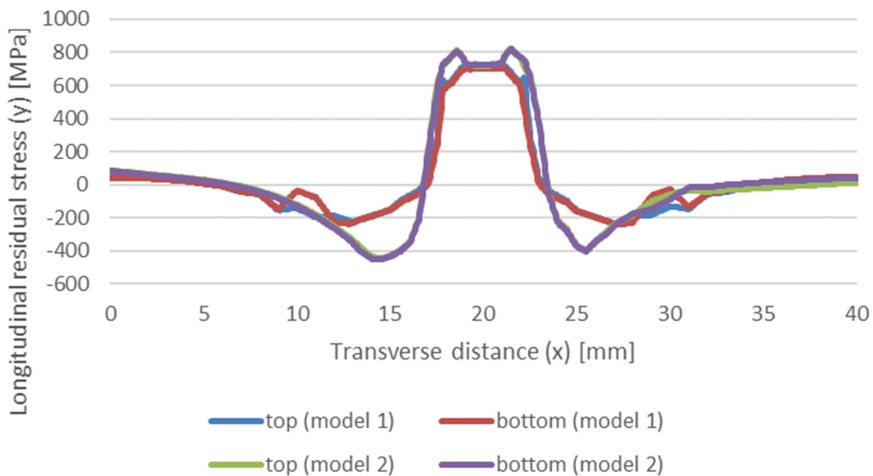


Figure 11.42 Comparison of longitudinal residual stress between models I and II.

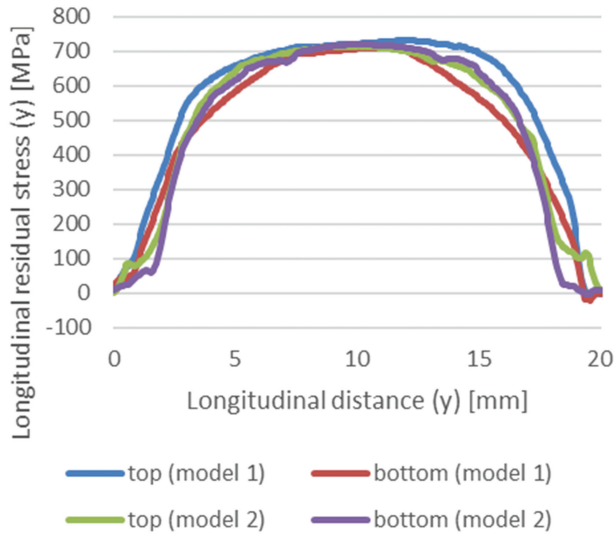


Figure 11.43 Comparison of longitudinal residual stress between models I and II.

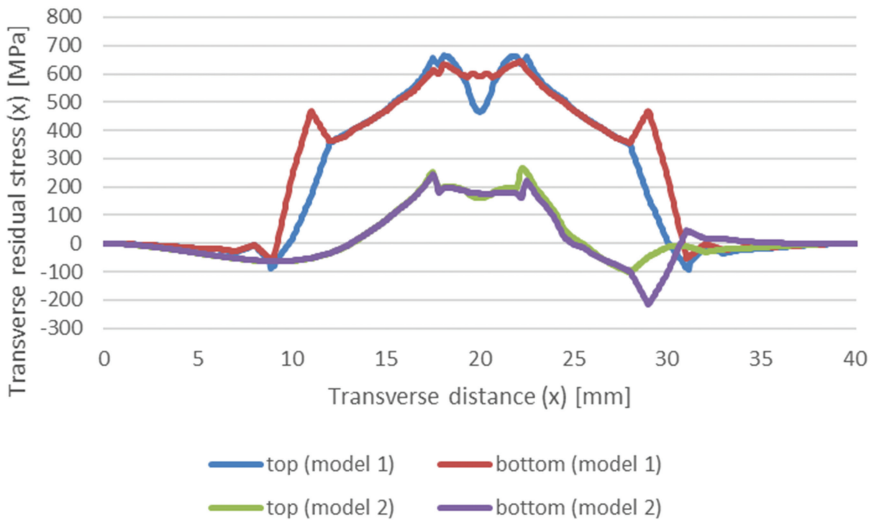


Figure 11.44 Comparison of transverse residual stress between models I and II.

Figure 11.45 presents the transverse stress along the longitudinal lines. At the plates' edges, a state of compressions is present with values around -600 MPa. This is verified for both models. In addition, it becomes clear that greater transverse stresses were predicted with model II. Also, model II has a considerable discrepancy between both top and bottom surfaces. The peak

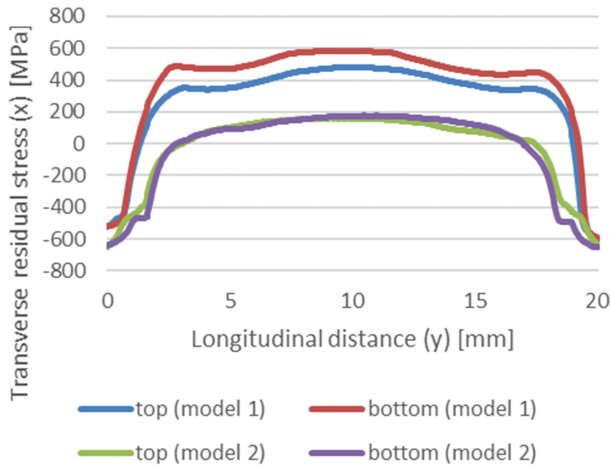


Figure 11.45 Comparison of transverse residual stress between models I and II.

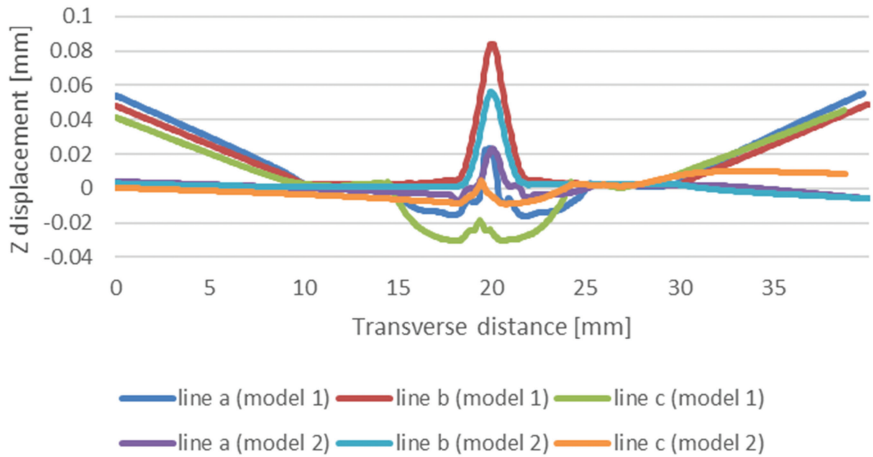


Figure 11.46 Comparison of Z-displacement between models I and II.

stress is approximately 600 MPa on the bottom surface. The difference in maximum stress between model II surfaces is around 100 MPa.

Model I presents a higher out-of-plane deformation, particularly at the edges of the plates far from the weld bead. Nevertheless, in the heat-affected zone, the behaviour of both models is similar. However, as depicted in Figure 11.46, the difference in magnitude increases with the welding direction.

Figure 11.47 shows higher predictions of the longitudinal deformation with model II. A justification is a rotational distortion due to the high welding

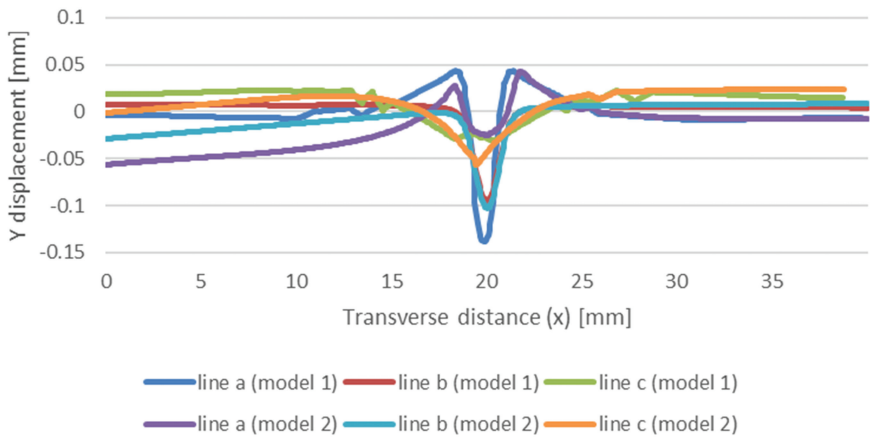


Figure 11.47 Comparison of Y-displacement between models I and II.

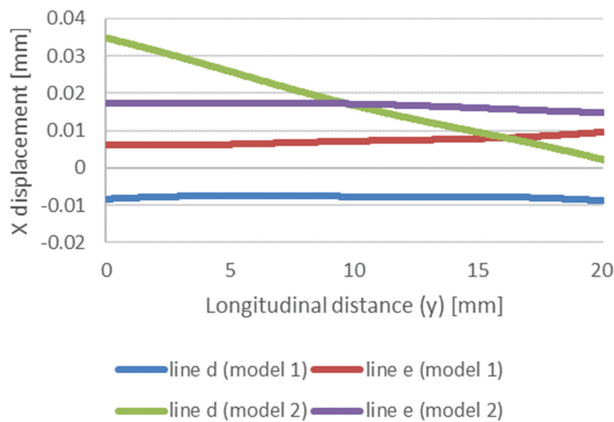


Figure 11.48 Comparison of X-displacement between models I and II.

speed. Regarding the transverse deformation, these are low, particularly with model I (Figure 11.48).

Conclusion

Two constraint configurations were analysed for laser welding of butt joints of 1 mm thick high-strength metal sheets (DP 600). Overall, the achieved numerical results are in agreement with the literature. The main conclusions, based on the models developed in this work (i.e., process parameters, materials, clamping configurations, etc.), are the following:

- Clamping conditions do not affect the longitudinal residual stress in this configuration.
- Clamping conditions significantly influence the transverse stress magnitude. Nevertheless, its profile is equivalent. The residual stresses are higher if both plates are constrained in the transverse direction.
- The out-of-plane deformation is low in both models, predicting higher values in the heat-affected zone. The deformation is nonuniform across the width.
- Models I and II have distinct predictions of longitudinal distortion. In model II, the edges have no deformation. In model I, there is a rotational distortion. Both models have longitudinal shrinkage near the bead.
- Transverse displacement is negligible in model II.
- In both cases, there is almost no difference in the residual stresses across the thickness. Same for deformation.

Funding

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Chapter 12

Prediction of Residual Stresses and Distortions of Laser Welded T-joints – A Numerical Approach

F. A. O. Fernandes^{1,2}, A. B. Pereira^{1,2} and M.C. Silva¹

¹TEMA—Centre for Mechanical Technology and Automation, Department of Mechanical Engineering, University of Aveiro, Campus de Santiago, 3810-193 Aveiro, Portugal

²LASI—Intelligent Systems Associate Laboratory, Portugal

Abstract

Welding is a joining process vastly employed to join several materials. Nevertheless, residual stresses and distortions are caused by the significant temperature gradient. The use of numerical methods to simulate the welding processes has been increasing significantly. Numerical simulation makes it possible to analyse complex phenomena and predict its final state. In this work, we developed a numerical model aiming to predict residual stresses and distortions of laser welded components in T-joint configuration. The modelling focus on the Dual-Phase 600 steel, highly employed in several industries and subjected to laser welding, e.g. automotive industry. The resulting distortions and stresses were quantified using the Simufact Welding.

Keywords: Laser welding, numerical simulation, Welding distortion analysis, DP600, Residual stresses, T-joint

Introduction

Welding results in distortions or residual stresses in the welded parts. The tremendous amount of concentrated heat promotes mechanical and metallurgical

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changes due to the significant temperature gradients typical in these processes. These transformations depend on many aspects, for instance, the material, process parameters and even clamping conditions.

The main cause of residual stresses in welding is usually linked to the significant thermal gradients and non-uniform thermal cycles, which may result in volumetric changes. These internal stresses exist without external forces applied (Nasir et al. 2016). Basically, the heating-cooling cycles cause material's expansion and contraction, resulting in states of compression-tension, originating the residual stresses after solidification (Schenk, n.d.). The prediction of internal stresses has many applications, such as the prediction of fracture resistance (Seang 2013).

Although vastly employed in several applications/industries, laser welding remains a relevant research topic. This is the case of the prediction of residual stresses and distortions due to the significant amount of concentrated energy that characterises the laser welding process. Nevertheless, the entire process is complex and highly nonlinear.

Numerical simulations make it possible to simulate complex multivariable problems. Numerical models are advantageous since these can be employed in designing, prototyping and optimisation, minimising the typically high number of experiments. Nevertheless, the simulation of laser welding is also complex. For example, the material suffers several transformations due to the heat generated from laser radiation (Kik 2020).

Welding processes simulation is characterised by its inherent complexity due to the nonlinearity of the phenomena (e.g., phase transitions) (Caprace et al. 2017). Dal and Fabbro (2016) reviewed laser welding simulation and highlighted the limitations of models usually employed to describe the heat source. Also, the need to account for the change of mechanical properties during welding is also highlighted and recommended for future model validations.

The simulation of welding processes usually can have many objectives but always aiming to improve the joint's quality. For instance, to study the welding parameters' influence on the weld pool (Klingler 2018) or to quantify the residual stresses and distortions (Fernandes and Pereira 2021). The elasto-plastic material behaviours strongly depend on the material's microstructure, which is affected and suffers transformations due to the significant thermal gradients. Knowledge of mechanical and thermal properties and how these can change during welding is necessary to define the model properly.

In this work, the aim is to predict residual stresses and distortions, particularly the vertical plate's maximum deformation of a laser-welded T-joint. The modelling focuses on the Dual-Phase 600 steel, highly employed in several industries and subjected to laser welding, e.g. automotive industry. The resulting distortions and stresses were determined using Simufact Welding.

Materials and Methods

This work aims to simulate Nd:YAG laser welding of T-joint. The residual stresses and distortions were determined by employing Simufact Welding.

Case Study – T-joint

The T-joint welding simulation is based on the laser Nd:YAG welding, with scanning speed and power of 3 m/min and 4 kW, respectively. Figure 12.1 illustrates the welding configuration, which employs 1 mm thick DP600 sheets, with a 20×41 mm rectangular plate placed horizontally and a 20 mm square vertical plate positioned in the middle of the horizontal one without a gap between them.

2.1.1. Numerical Setup

Each model has two steps: welding and cooling. The clamping system consists of rigid bodies with the following dimensions: $20 \text{ mm} \times 5 \text{ mm} \times 1 \text{ mm}$. Pairs of bearings-clampings were employed to constrain the plates during the welding. The bearing one (red in Figure 12.2) allows sliding on the same contact plane. The clamping system (green in Figure 12.2) presses the plates against the bearings with a 1 kN force and a constant translation stiffness of 5 kN/m.

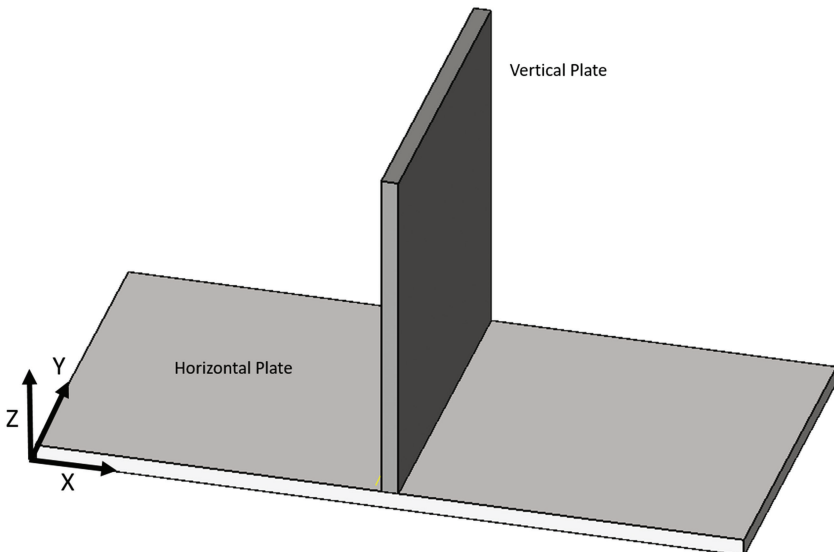


Figure 12.1 Illustration of the joint configuration.

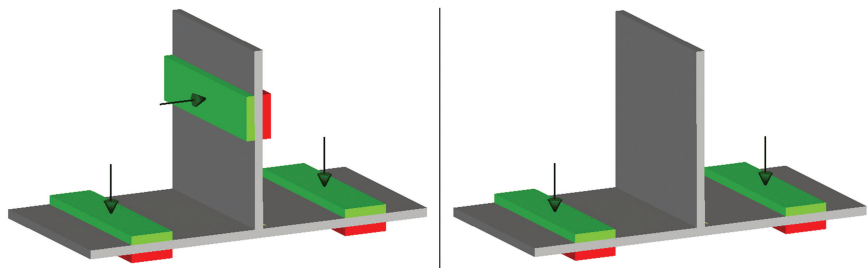


Figure 12.2 Clamping conditions at: (left) step A; (right) steps B and C.

Table 12.1 Clamping conditions

Step		Horizontal plate		Vertical plate	
		Clamp	Bearing	Clamp	Bearing
A	Welding (tack)	ON	ON	ON	ON
B	Welding (joining)	ON	ON	OFF	OFF
C	Cooling	ON	ON	OFF	OFF

Table 12.2 Steps control

Parameter	Welding step	Cooling step
Mechanical solution every n thermal increment	3	5
Result output every n increment	6	5

The centre lines of these boundary conditions distance approximately 12.5 mm from the bead’s centre line in the transverse direction. Since the aim is to predict the vertical plate’s maximum distortion, this is not constrained during welding. Table 12.1 summarises the changes in the boundary condition during the entire simulation.

Room temperature was set to 20 °C, and gravity was considered (9.80665 m/s²). The cooling step is activated when the welding is over and only ends when the workpiece decreases to 300 °C. This simplification has a negligible effect on the results and minimises the computational time necessary that would take to achieve room temperature.

The analysis was performed in precision mode, and both thermal and mechanical analyses were performed together. Phase transformations were disregarded. The total processing time for all models was 35 seconds to ensure the cooldown phase after the end of the welding phase. Table 12.2 presents the values employed to define the frequency of mechanical calculation relative to

thermal analysis and the results of output frequency for welding and cooling cycles.

Heat Source

Figure 12.3 depicts the conical heat source with Gaussian distribution in key-hole mode employed to simulate the laser welding. Table 12.3 presents the dimensions of the heat source, characterised by uniform power distribution.

Figure 12.4 shows the trajectory of the heat source. The weld bead has a total length of 20 mm. Two tack welds were made, each 1 mm long.

Material

The material properties available in the material library of Simufact Welding (S60QL-SPM-sw) and with similar properties to DP600 steel were employed. The initial phase fractions assume 79% bainite and 21% ferrite. The material presents a solidus and melting temperatures of 1462.92 °C and 1509.95 °C, respectively. The latent heat for melting has a value of 256.4 kJ/kg. The thermal conductivity is temperature-dependent, as described in Figure 12.5. Figure 12.6 depicts the material’s specific heat capacity thermal dependency.

Regarding the mechanical properties, the Poisson’s ratio was defined with a constant value of 0.3. The Young’s modulus temperature dependency is depicted in Figure 12.7.

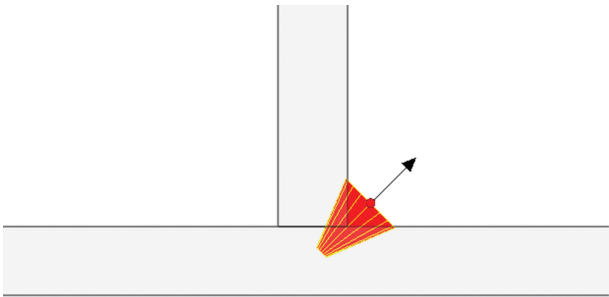


Figure 12.3 Heat source.

Table 12.3 Heat source dimensions

Parameter	T-joint [mm]
Radius superior (r_u)	0.50
Radius inferior (r_i)	0.10
Depth (d)	1.00
Offset	0.47

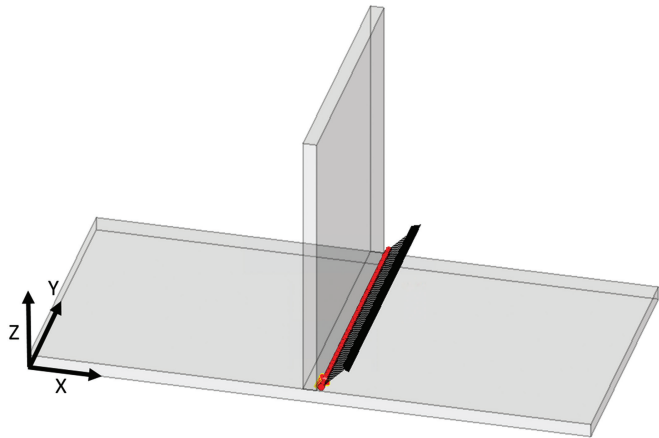


Figure 12.4 Heat source trajectory.

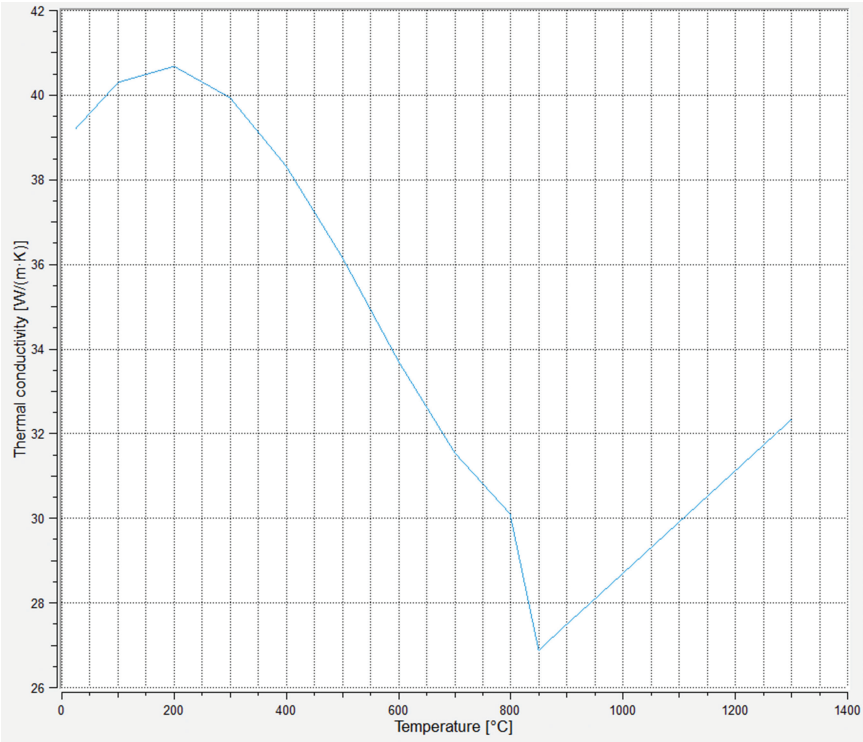


Figure 12.5 Thermal conductivity vs temperature.

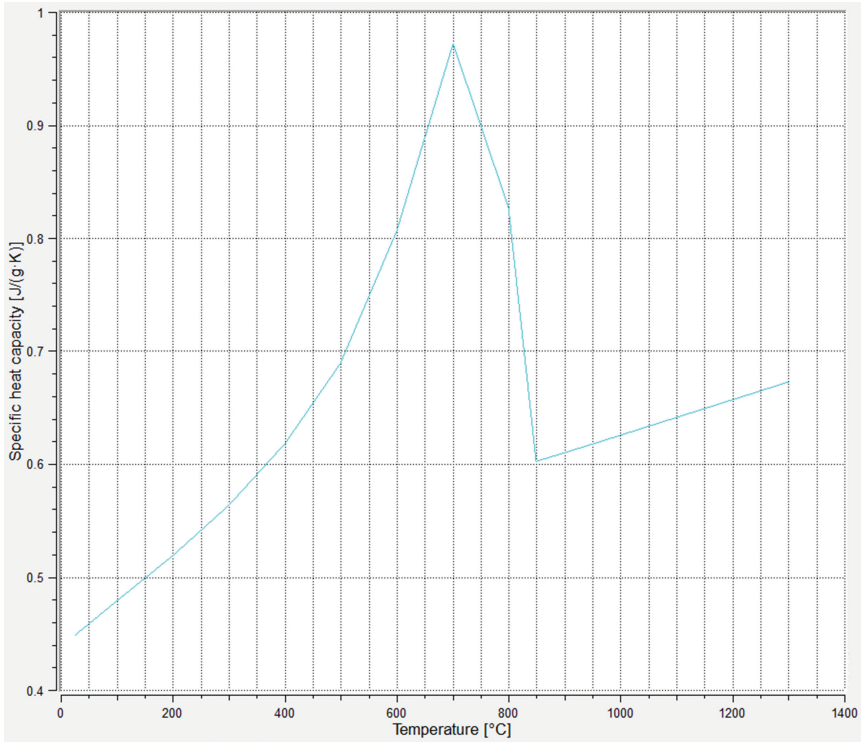


Figure 12.6 Specific heat capacity vs temperature.

Figure 12.8 depicts the density dependency on the temperature, while Figure 12.9 shows the temperature dependency of the thermal expansion coefficient, which describes the material's volumetric changes with temperature.

Figure 12.10 shows the flow curves at different temperatures (25 °C, 100 °C, 200 °C, 300 °C, 400 °C, 500 °C, 600 °C, 700 °C, 800 °C, 900 °C, 1000 °C, 1100 °C, 1200 °C and 1300 °C), necessary to define the true stress–true plastic strain relationship for several temperatures. As the temperature increases, the lower the stress level is.

Mesh

A mesh convergence study was conducted to mitigate the mesh influence on the results and to avoid excessive mesh refinement, which would result in excessive computational time. The final model is sectioned into three regions, from less to more refined mesh: base material, heat affected zone and fusion zone. The final model is composed of 17956 hexahedral finite elements with 3 elements

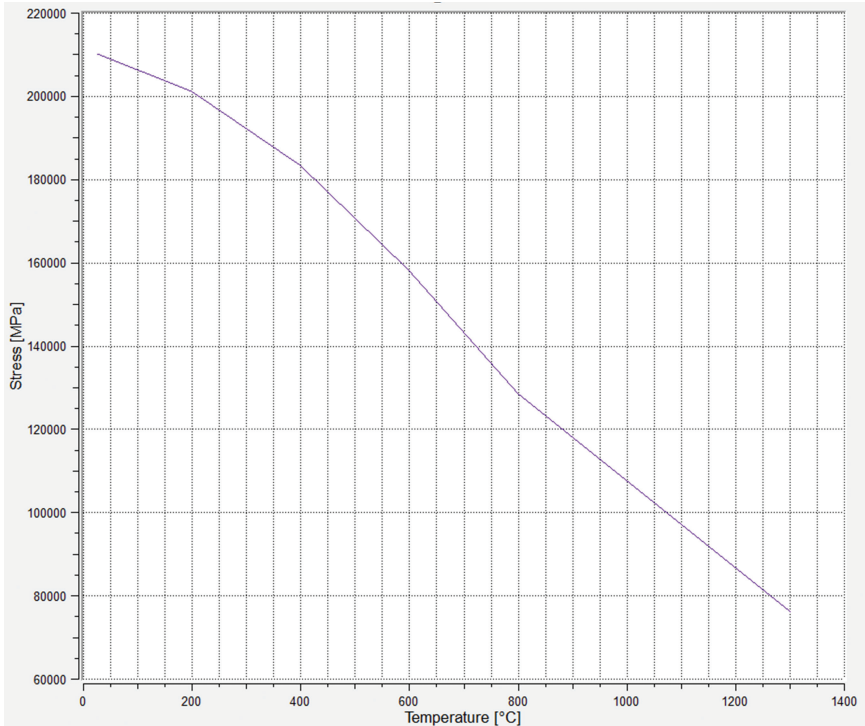


Figure 12.7 Young’s modulus vs temperature.

across the thickness, and 8160 nodes (Figure 12.11). The elements’ size ranges from 0.25 mm × 0.25 mm × 0.33 mm to 1.0 mm × 1.0 mm × 0.33 mm.

Results and Discussion

Residual Stress Measurement

Different transverse lines are defined to monitor both plates’ residual stress and distortions. Nevertheless, as illustrated in Figure 12.12, the longitudinal line on the bead is the same for both plates. In the vertical one, there are two lines to measure the transverse and stresses. In this one, all lines are 20 mm long with 40 tracking points per millimetre.

Figure 12.13 presents another view highlighting new lines on the vertical plate: right and left surface (green and purple, respectively). There is a transverse right line and a longitudinal one on the right surface. It is on this side

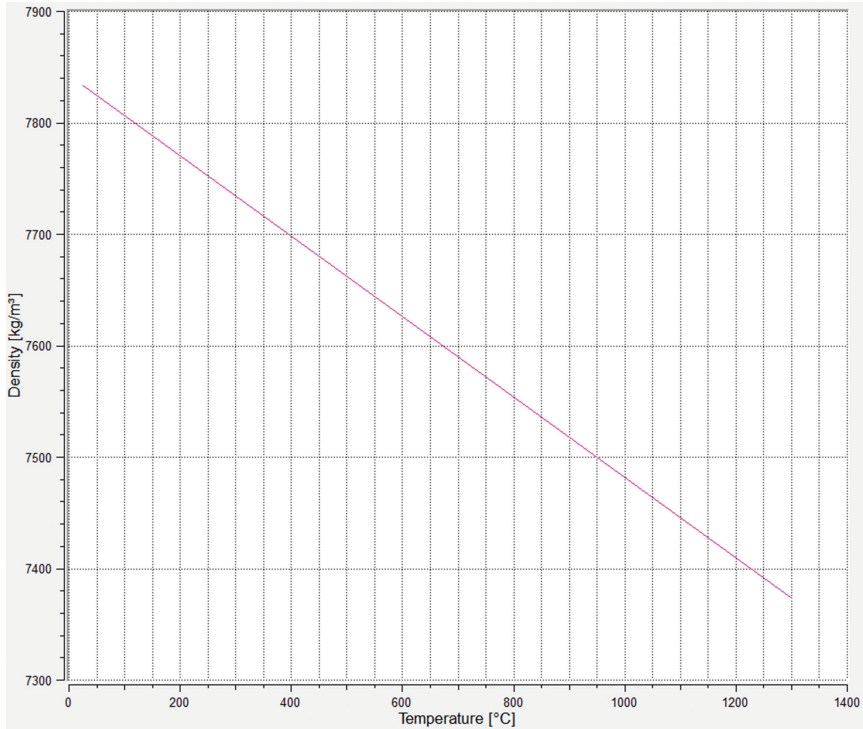


Figure 12.8 Density vs temperature.

that the laser beam will interact. On the opposite side, there is only one line (purple).

In the horizontal plate, two transverse and two longitudinal lines are used to determine the transverse and longitudinal stresses, respectively (Figure 12.14). Two are positioned on the top surface (orange), and the other two are on the bottom surface (green) (Figure 12.15). The single difference is the 41 mm long transverse lines to account for the vertical plate thickness.

Displacement Measurement

Three lines are defined only on the side of the bead (right surface) to determine the vertical plate's x-displacement (Figure 12.16). This approach of disregarding the other side is reasonable due to the low thickness. Figure 12.17 shows the position of the longitudinal lines to determine the Z-displacement.

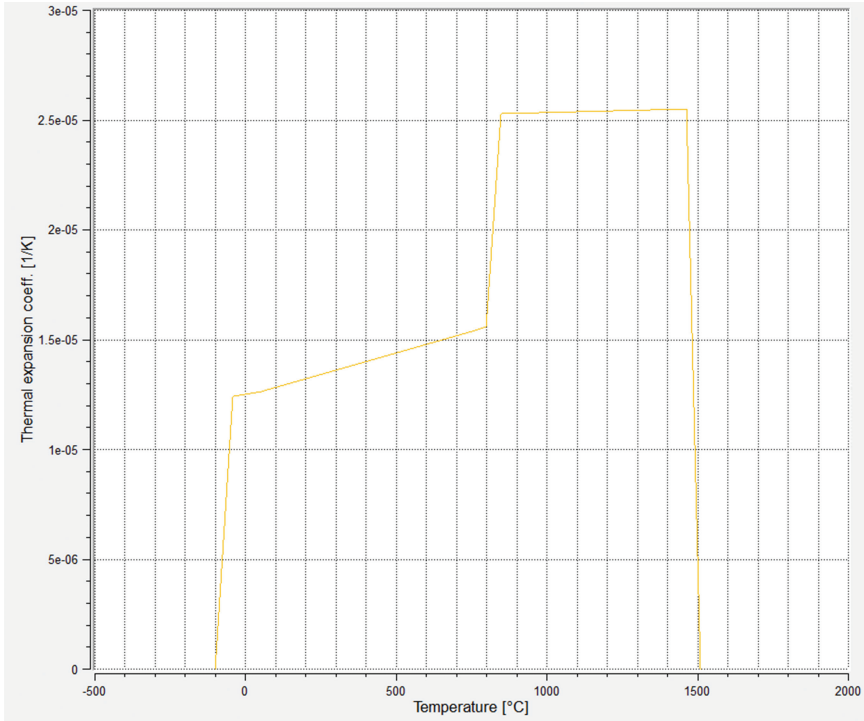


Figure 12.9 Thermal expansion coefficient vs temperature.

Figure 12.18 shows the 3 transverse lines (a, b and c) and the 2 longitudinal lines (d and e) defined to determine the horizontal plate's distortion. The 41 mm long transverse lines are equidistant, while the longitudinal ones are 20 mm long.

Results – T-Joint Case Study

Transverse Stress

Figure 12.19 shows a higher transverse stress concentration at the vertical plate's lower part. Nevertheless, these stresses are slightly lower in the plate's centre.

Figure 12.20 shows a minimal influence of the residual stress along vertical plate's thickness. Also, it is clear the high stress near the bead. For $0 \leq z \leq 1$ mm, there is a compression state reaching -70 MPa. For $1 < z \leq 3$ mm there is a state of tension, followed by another compression one with a peak value under -100 MPa, which reduces until 0 MPa (vertical plate's top).

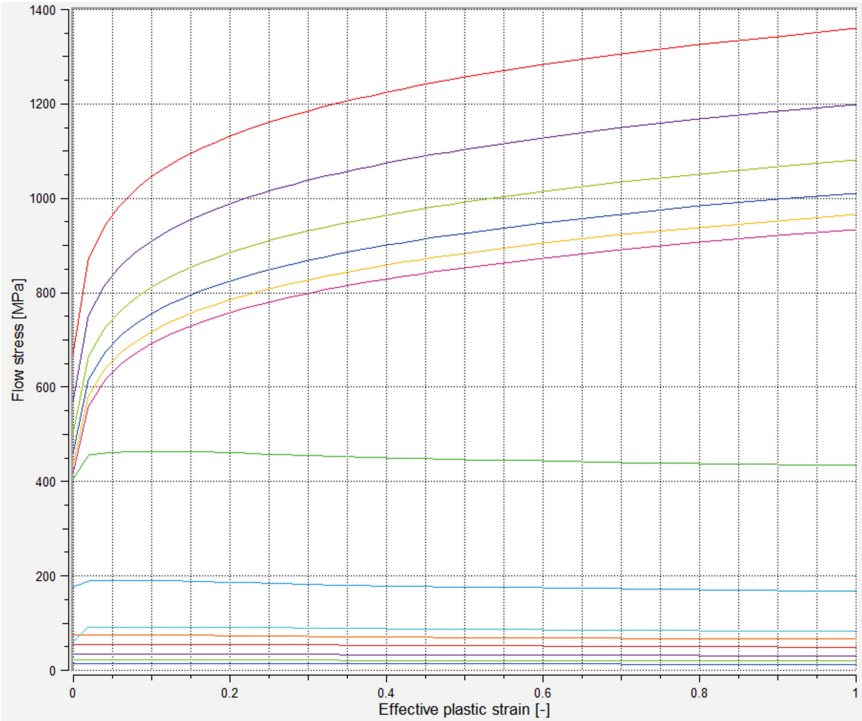


Figure 12.10 Flow curves (the stress levels decrease with temperature evolution).

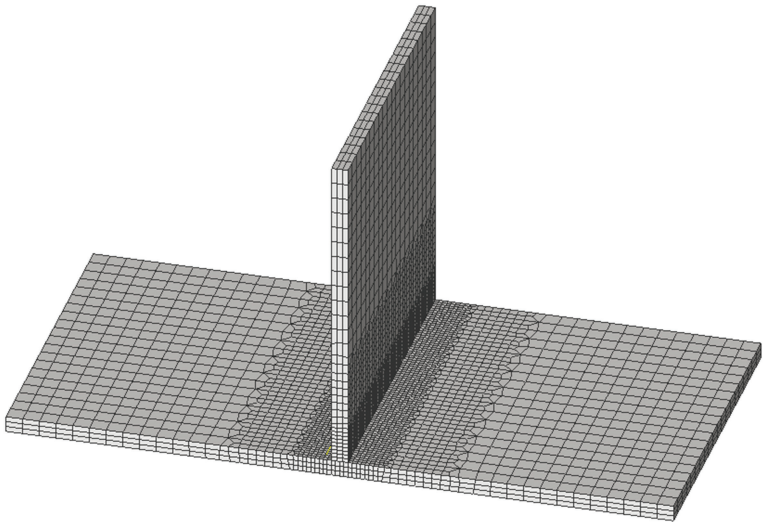


Figure 12.11 Final mesh obtained after the convergence study.

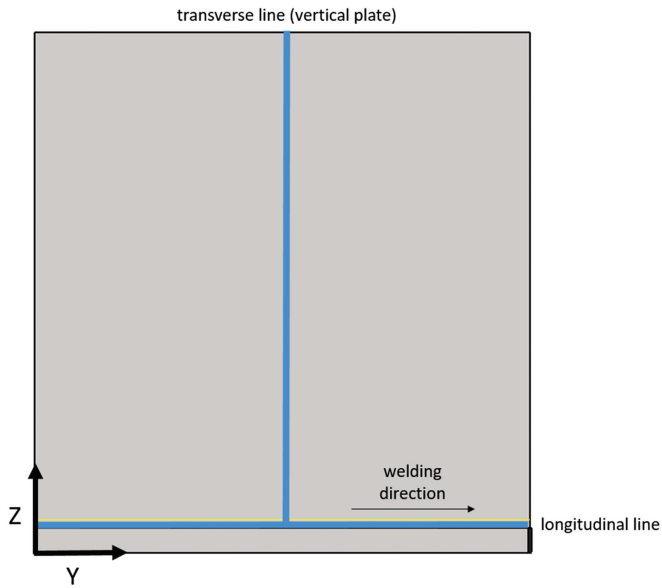


Figure 12.12 Transverse and longitudinal lines on the vertical plate.

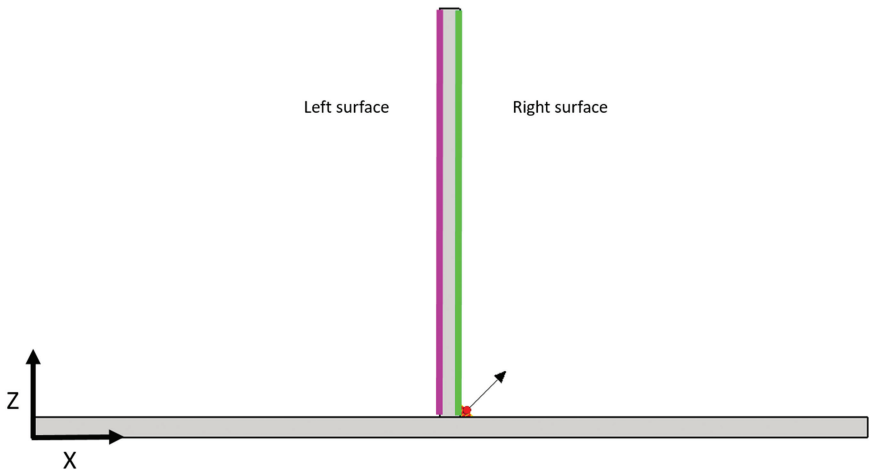


Figure 12.13 Additional lines to monitor vertical plate.

Figure 12.21 highlights the transverse stresses present in the horizontal plate (x-direction). Two regions of interest are spotted (higher tensile stress concentration): edges and bead’s centre.

Figure 12.22 shows positive stresses with a peak value of around 350 MPa and located in the central region of the bead. No relevant differences are

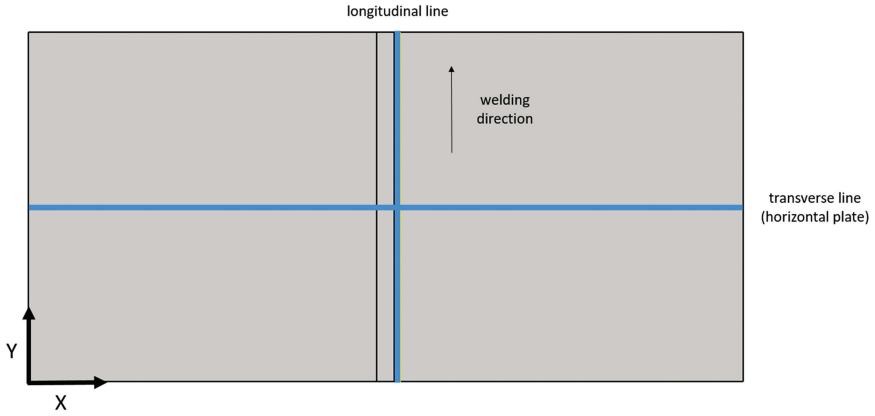


Figure 12.14 Transverse and longitudinal lines of the horizontal plate.

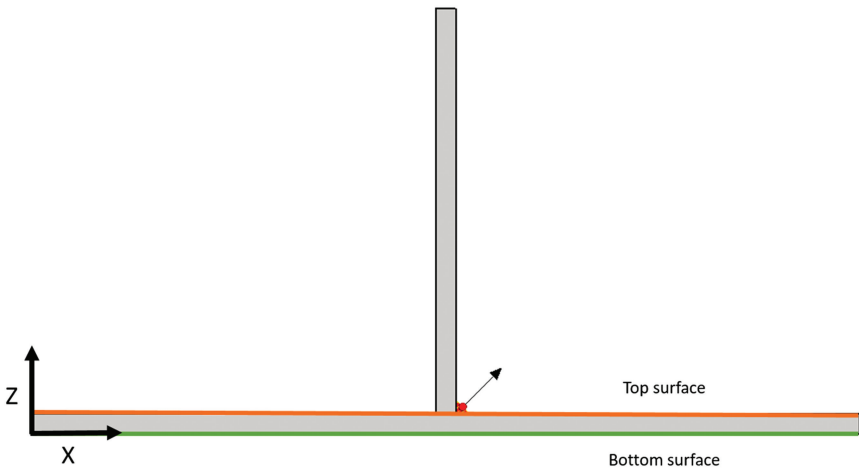


Figure 12.15 Top and bottom surfaces.

observed between the cross-thickness stress magnitudes. The stress is null at the edges of the horizontal plate, increasing symmetrically up to joint.

Figure 12.23 depicts the transverse stress along the longitudinal line, showing significant contrast between vertical and horizontal plates. Nevertheless, symmetry is observed in both cases. The stress gradient is much more significant in the horizontal plate, from a compression state of -700 MPa to a tensile one reaching 200 MPa. In the transverse direction, the vertical plate is practically free from residual stress, except near the edges, where a compression state is observed.

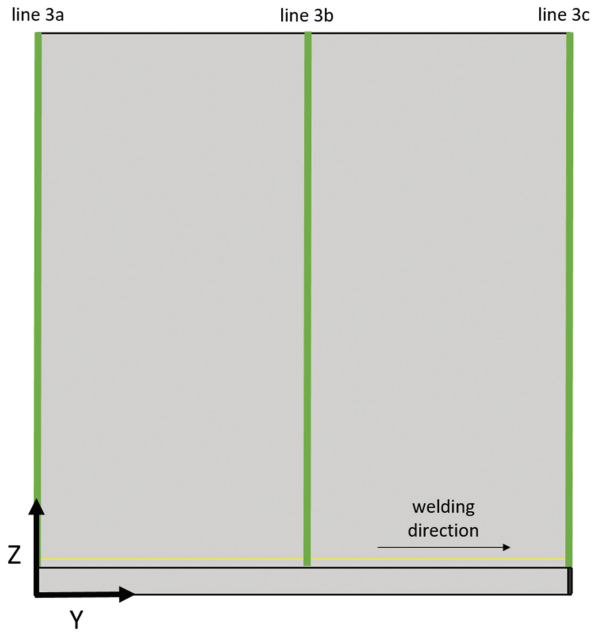


Figure 12.16 Transverse lines to determine X- and Y-displacements.

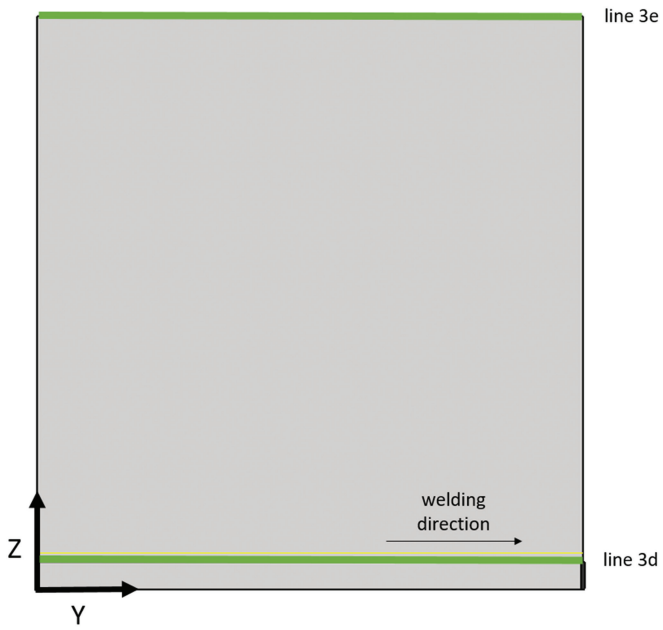


Figure 12.17 Longitudinal lines to determine the Z-displacement.

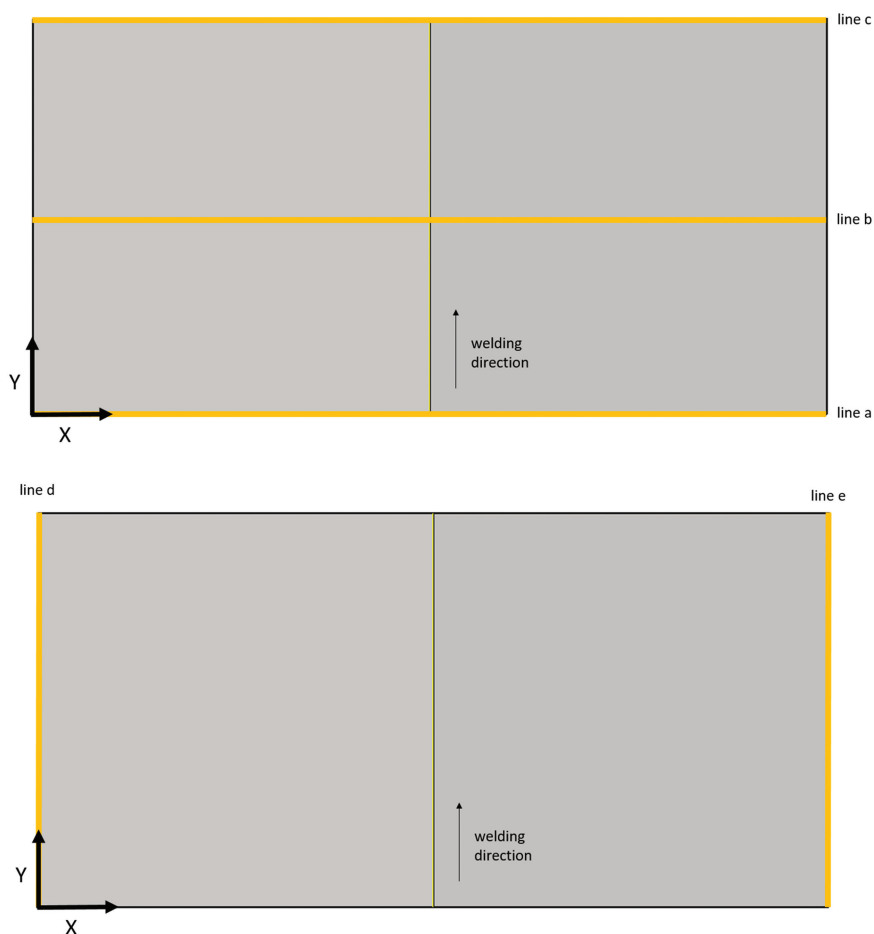


Figure 12.18 (Left) Transverse lines to determine Y- and Z-displacements; (right) longitudinal lines to determine X-displacement.

Longitudinal Stress

Figure 12.24 reveals the region with higher stress concentration, which is close to the bead. Figure 12.25 shows that the longitudinal stress along the transverse lines of the horizontal plate is practically identical and symmetrical. A compression state with a maximum value of approximately -400 MPa is observed close to the bead region. On the other hand, a tension state with a peak value of around 800 MPa is observed in the plate's middle.

Figure 12.26 shows a tension state close to 800 MPa at the vertical plate, close to the bead ($y = 0$ mm). For $y = 3$ mm, there is a transition to a state of

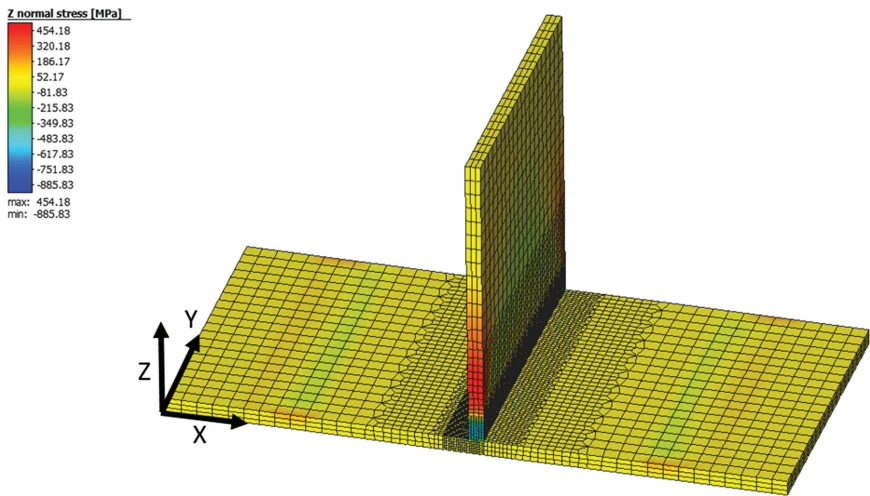


Figure 12.19 Residual transverse stress mapping (z-direction).

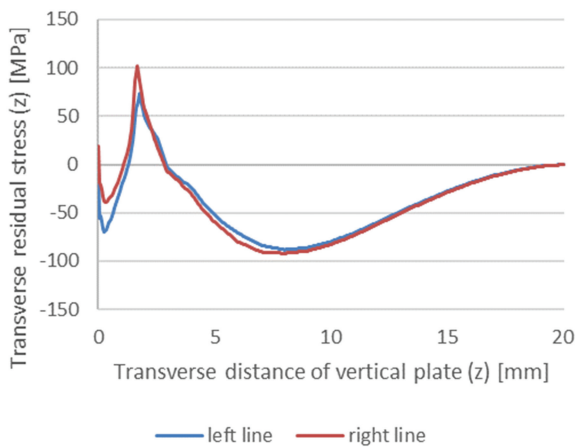


Figure 12.20 Transverse stress along the vertical plate’s transverse lines (origin corresponds to the bead).

compression. The lowest longitudinal residual stress value is approximately -400 MPa.

Figure 12.27 presents the longitudinal stress based on the longitudinal line, common to both plates. It is clear that there is symmetry. It is also noteworthy that a tensile state with a maximum residual stress value of around 740 MPa.

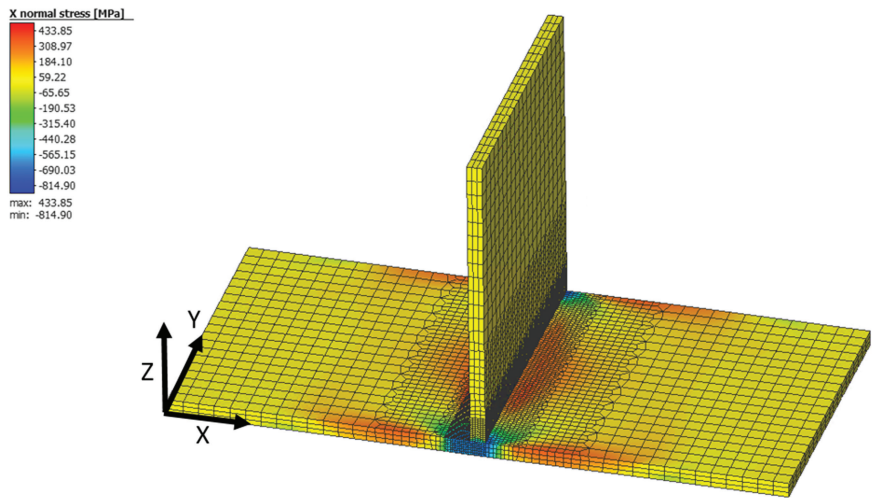


Figure 12.21 Transverse residual stress mapping (x-direction).

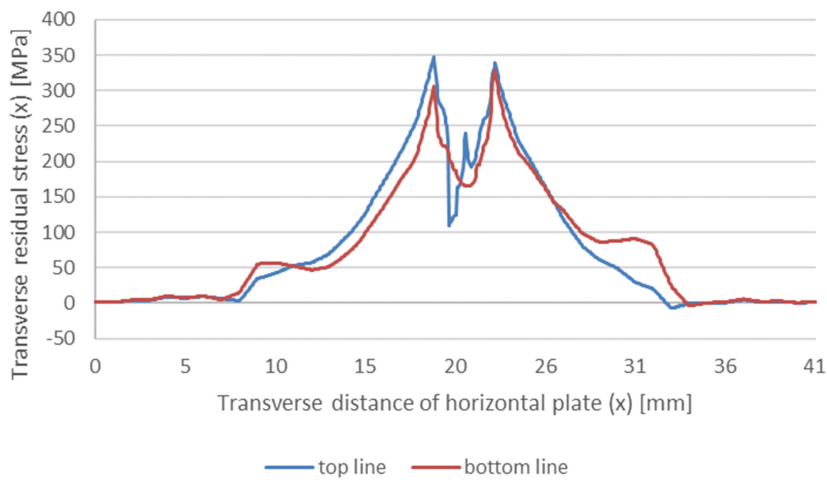


Figure 12.22 Transverse stress along the transverse lines (horizontal plate).

X-displacement

Since the vertical plate is unconstrained during welding and the vertical one is clamped, the x-deformation is much higher on this plate, as shown in Figure 12.28. Figure 12.29 depicts the angular distortion between both plates.

Figure 12.30 shows the linear deformation of the vertical plate. The average maximum displacement of the three lines is 0.26 mm (top of the plate).

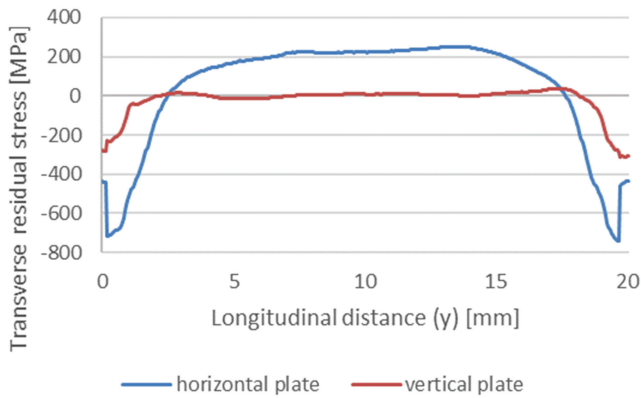


Figure 12.23 Transverse stress along the longitudinal line.

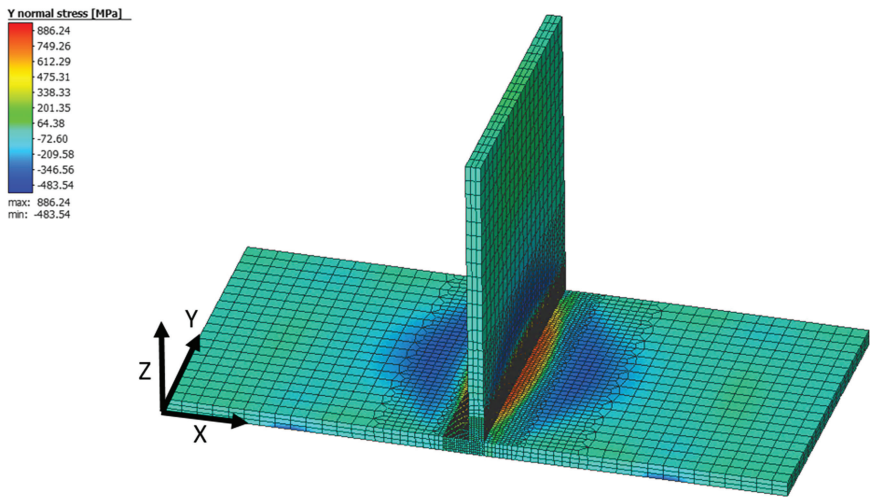


Figure 12.24 Longitudinal residual stress mapping (y-direction).

Considering the plate’s length, it has approximately 1.6 ° of angular distortion. Bending occurs on the bead side.

In Figure 12.31, it is possible to observe the scissor effect – positive rotational distortion in the horizontal plate. Thus, the distance between the horizontal plate edges tends to decrease. There is a symmetrical behaviour in relation to the joint.

Y-displacement

Figure 12.32 shows that the y-displacement is basically null for both plates, except for the bead. Figure 12.33 shows a higher shrinkage close to the bead

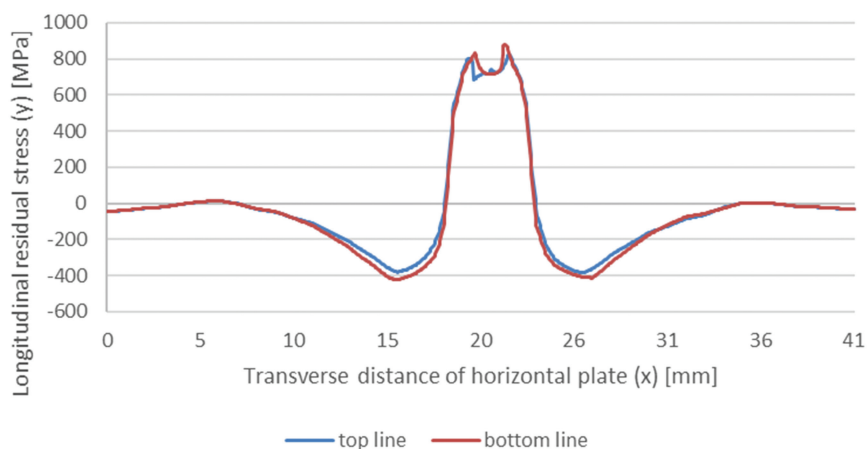


Figure 12.25 Longitudinal stress along transverse lines of the horizontal plate.

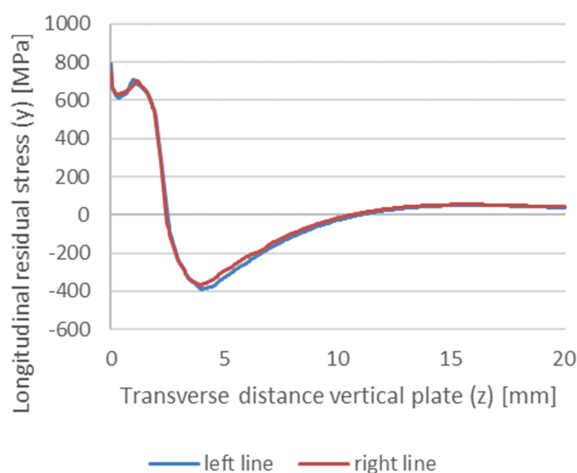


Figure 12.26 Vertical plate's longitudinal stress along transverse lines.

of ± 0.02 mm (lines b and c). There is a higher y-displacement of -0.1 mm in the joint, which considering the dimensions of the sample, it is a negligible deformation,

Figure 12.34 reveals the existence of small transverse shrinkage for the vertical plate. Nevertheless, it is not uniform along the plate. At the beginning (line a), middle of the plate (line b) and end of the plate (line c), the displacements are of around -0.04 mm, -0.02 mm and almost null, respectively. Regarding the horizontal plate, a significant displacement in the bead region goes up -0.1 mm.

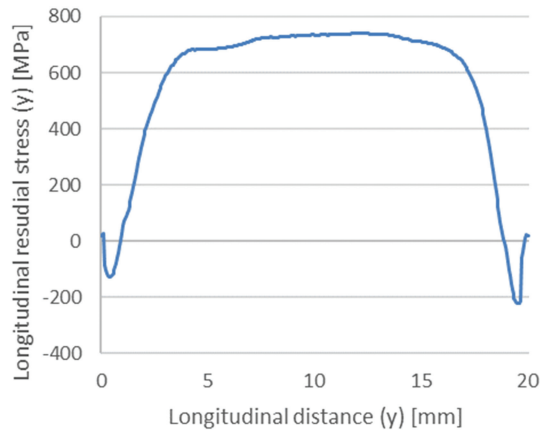


Figure 12.27 Longitudinal stress along the longitudinal line.

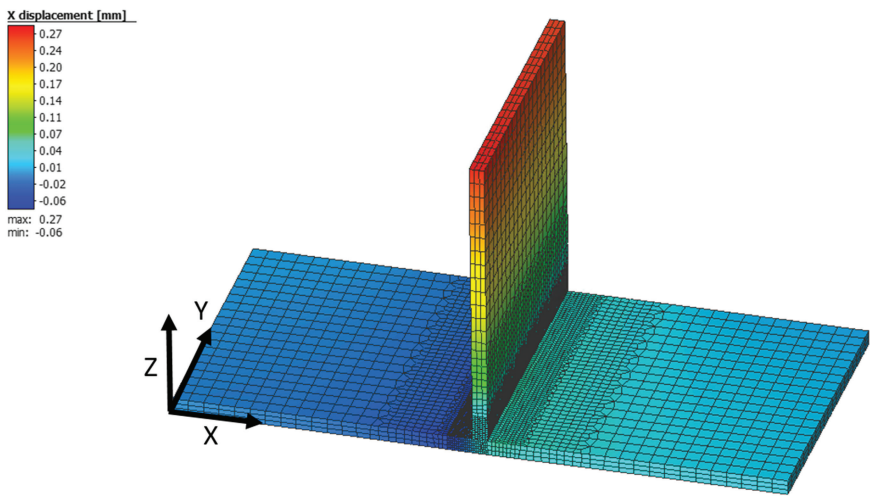


Figure 12.28 Overview – X-displacement.

Z-displacement

Figure 12.35 presents the z-displacement, which is null in the entire welded component, except for the bead region. Outside the weld bead zone, this displacement is practically zero.

Figure 12.36 depicts the horizontal plate’ out-of-plane displacement. Null out-of-plane displacements are measured in the clamping areas. Two main behaviours characterise the out-of-plane displacement between the

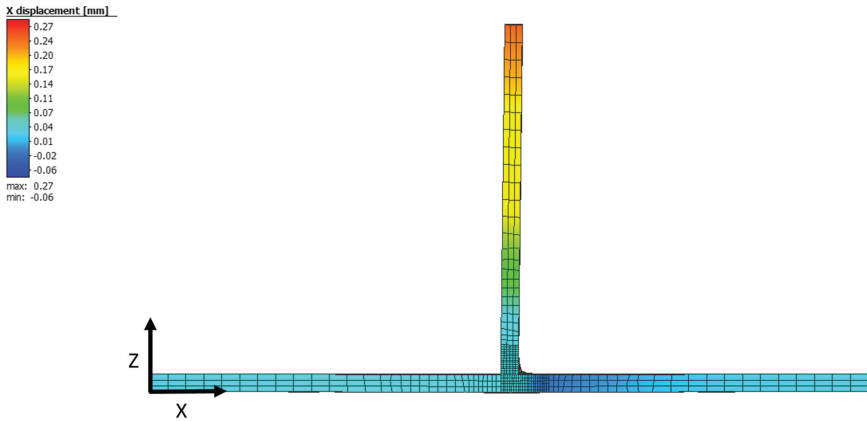


Figure 12.29 Side view – X-displacement.

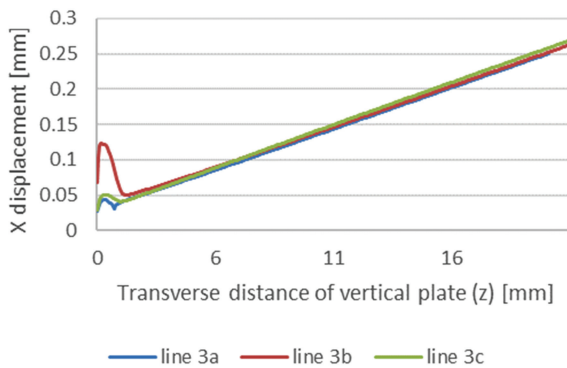


Figure 12.30 Vertical plate's x-displacement along transverse lines.

clamps and the bead. There is a positive out-of-plane displacement in the central line b, reaching a maximum value of 0.08 mm. At the edges, the out-of-plane displacement is negative, reaching a maximum magnitude value over -0.04 mm.

Since this displacement has the potential to influence the vertical plate's length, it was also verified. From the results depicted in Figure 12.37, one can conclude that the changes in the length of the vertical plate are negligible. The line positioned at the base of the vertical plate (line d) presents a different behaviour than line e. The maximum displacement is measured at the edges, over 0.08 mm. A symmetric behaviour is also observed for an x–y plane at $y = 10$ mm.

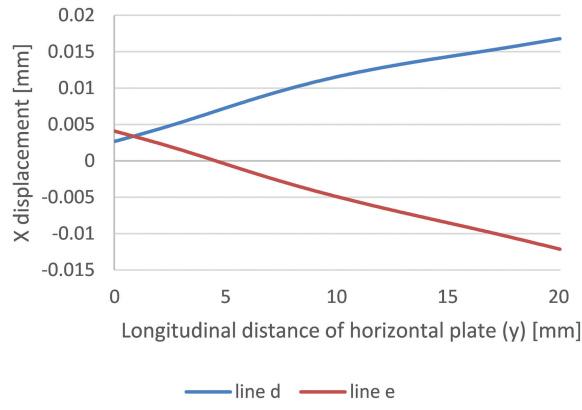


Figure 12.31 X displacement along longitudinal lines of the horizontal plate.

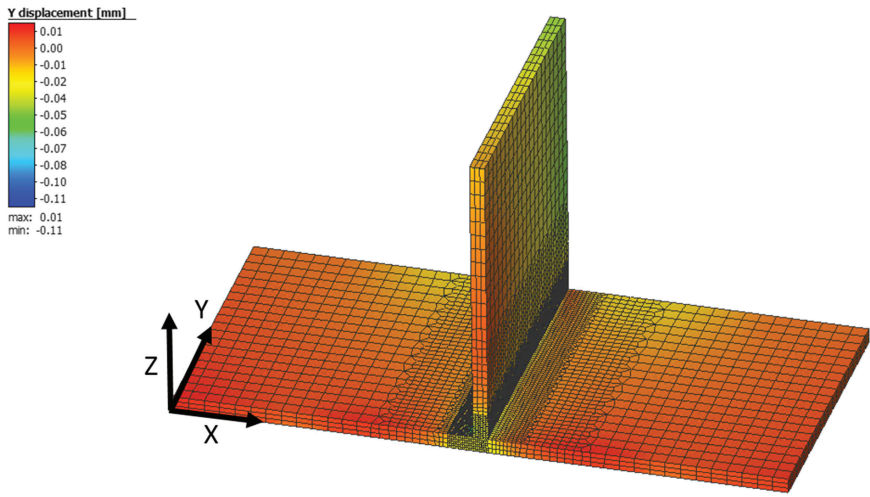


Figure 12.32 Overview – y-displacement.

Discussion and Conclusion

Predicting the final state of a welded component is a complex task. Numerical simulations are a powerful tool for predicting such phenomena. Overall, the main findings of this work were the following:

- Regarding the transverse stress:
 - a) Regarding the vertical plate, there is a higher concentration of residual stress near the bead at the transverse lines. Starts with a small

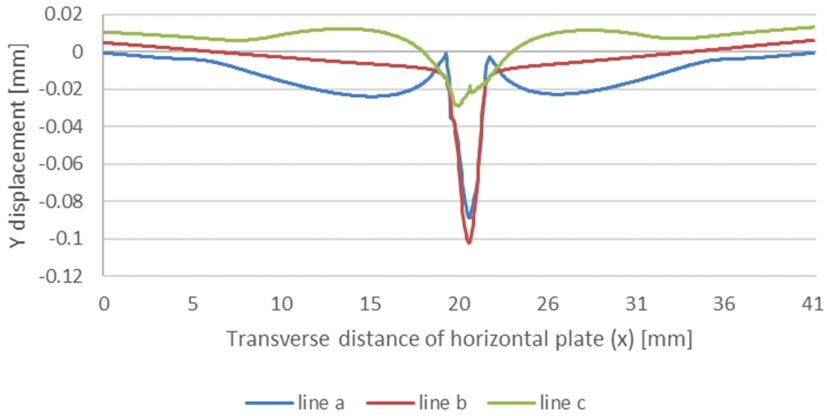


Figure 12.33 Horizontal plate's y-displacement along the transverse lines.

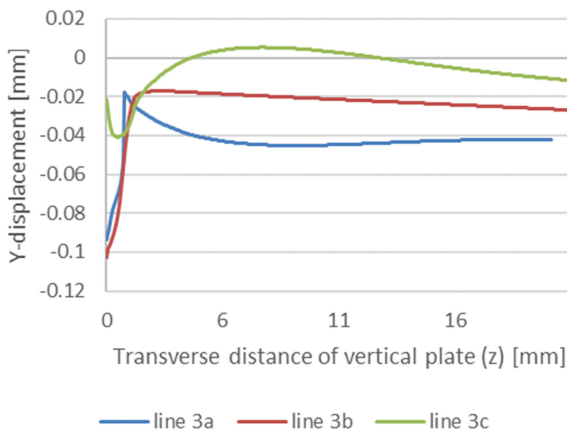


Figure 12.34 Vertical plate's y-displacement along transverse lines.

- compression state transiting to a tensile one and returning to a compression one.
- b) For the same lines and regarding the horizontal plate, the residual stress profile is symmetrical and always positive.
- c) The prediction of the horizontal and vertical plates' residual stress along the longitudinal lines highly differs from each other.
- Regarding the longitudinal stress:
 - a) There is symmetry along the transverse lines for the horizontal plate. There is a tensile state with peak values measured next to the bead.

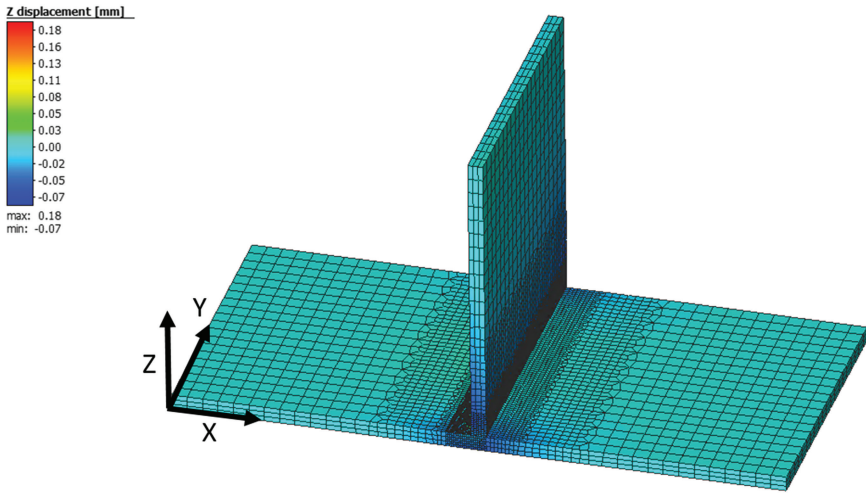


Figure 12.35 Overview – z-displacement.

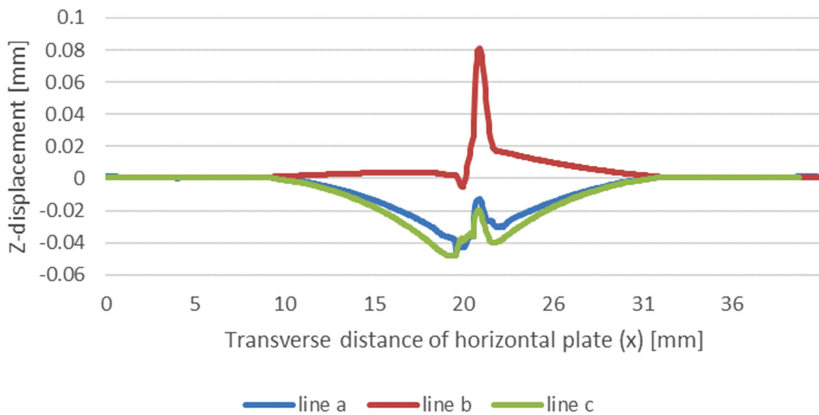


Figure 12.36 Horizontal plate's out-of-plane displacement along the transverse lines.

- b) Regarding the vertical plate, residual stresses are close to the bead for the same lines.
- c) About the measurements along the longitudinal line, both tensile and compression states are observed, the first at the edges and the second at other regions.
- There is an angular distortion under 2° registered for the vertical plate. There is a slight scissor effect at the edges of the horizontal plate.
- A small transverse shrinkage in the y-direction was also observed along the transverse edges, which was also verified for the vertical plate.

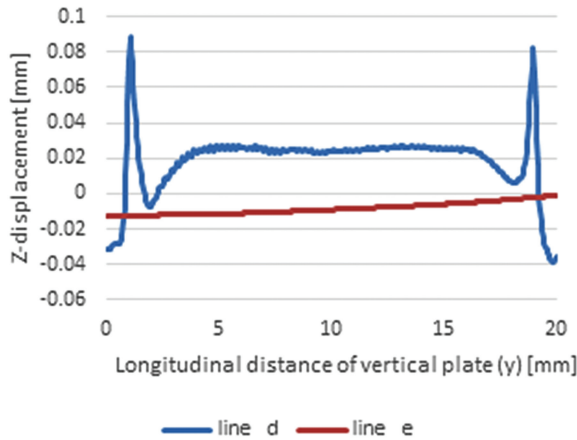


Figure 12.37 Vertical plate's z-displacement along longitudinal lines.

- The z-displacement is practically zero.
- The residual stress magnitude difference and deformation across the thickness are both negligible.

Overall, the model developed and the results computed with it agree with the literature.

Funding

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Chapter 13

Monitoring and Quality Assurance of Laser Welding: From Offline Sample-Based Testing to In-Process Real-Time AI Inference and Digital Twins

**P. Stavropoulos¹, K. Sabatakakis¹, H. Bikas¹ and
A. Papacharalampopoulos¹**

¹ Laboratory for Manufacturing Systems and Automation, Department of Mechanical Engineering and Aeronautics, University of Patras, Patras, Greece

Abstract

Flexible processes such as laser welding have been extended and established in modern manufacturing as they pose several advantages compared to conventional welding methods such as high productivity, flexibility, effectiveness, and numerous more such as deeper penetration, lower distortions, and high welding speeds. Despite that, laser welding remains a complex manufacturing process for which the prediction or even the visually recognizable quality of the weld is affected by several variables and factors, such as defects in the material's microstructure, contaminations on the workpiece surface, and alteration on the laser beam's properties, resulting in a not acceptable product. Consequently, monitoring and quality control systems are critical and necessary tools for welding results to be kept within deterministic boundaries and deal with changing conditions without the need for demanding and time-consuming manual setup. To this end, Data-driven and physics-based modeling has been the core of feature prediction and quality assurance approaches for laser welding. By utilizing sensor interfaces for capturing the process emissions or based on post-process inspection these models along with the process input data are engaged within the context of edge monitoring systems either decentralized or connected to Enterprise

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Resource Planning platforms. In addition, state-of-the-art approaches are introducing process feedback and validation by introducing relevant models into Digital Twins and Cyber-physical Systems.

Keywords: Laser welding monitoring, Artificial Intelligence, Digital Twins

Introduction

Among manufacturing processes, lasers are one of the major contributors to increased flexibility (Chryssolouris 2013). When it comes to welding application Laser Welding has seen an increased adaptation, especially in the Automotive industry as the welding process can be easily automated by using Computer-Aided Design/Manufacturing tools/setup (Ceglarek et al. 2015). Laser Welding is used for joining applications that require high processing speed and precision joining of dissimilar materials, and for parts that cannot be subjected to post-processing (e.g., grinding of the welded bead) or deformed easily by the excessive heating of conventional welding processes (Wang et al. 2016).

In welding defects are these imperfections that can be caused by a bad process/weld design but also by uncontrollable external (to the process environment) factors. Despite the several advantages, Laser Welding also presents several challenges as its working principles are based on complex mechanisms either for conduction, transition, or keyhole mode. Achieving the desired weld bead profile is not trivial and depends on various parameters such as the power density, interaction time, and specific point energy which depend on process parameters such as the power, speed, laser spot size, and power distribution profiles (Suder and Williams 2012). This is because different combinations of process parameters may produce for example the desired penetration depth, but some of them may also cause defects such as expulsion or porosity (Hann, Iammi, and Folkes 2011). High reflective metals (e.g., aluminum, copper, gold) have low energy coupling efficiency with the energy absorption made mainly at the lower part of the keyhole due to multiple reflections (Ning et al. 2019). This makes the keyhole unstable as the energy distribution is not homogenous and eventually this could cause defects such as porosity. Keyhole instability and collapsing however is kind of a universal problem in deep penetration Laser Welding at low speeds and when the based metals are coated with metals of lower vaporization points (Katayama, Kawahito, and Mizutani 2010; Lin et al. 2017). Besides, porosity defects like humping can be also due to molten pool instability which appears mainly when the length of the pool is getting bigger, and the welding speed is high (Ai et al. 2020). Cracking is another

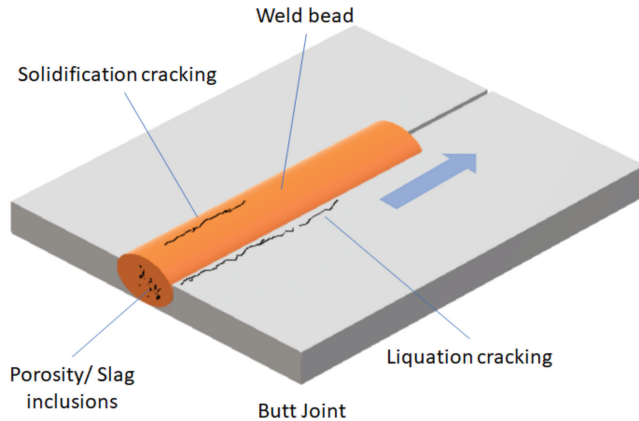


Figure 13.1 Most common defects in Laser Welding: porosity, solidification, and liquation cracking

common defect in Laser Welding. Solidification cracks start in the mushy zone during the last stage of solidification as a result of shrinkage, thermal stress/strains, hindered contraction, and lack of liquid feeding (Sheikhi, Ghaini, and Assadi 2015), while liquation cracking occurs in the partially melted zone by the liquid film and the subsequent de-cohesion along the inter-granular solid-liquid interface under the tensile stress and concerns mostly the welding of dissimilar metals and superalloys (Li et al. 2020; Pakniat, Ghaini, and Torkamany 2016). Beyond these defects (Figure 13.1) which are particularly common for Laser Welding, other defects such as lack of fusion, evaporation & loss of alloy elements at the fusion zone, burn-through, and bad appearance can be also observed for Laser Welding (Katayama 2013).

To this end strategies for controlling features of the weld bead profile (Hann, Iammi, and Folkes 2011; Kos et al. 2021) and limiting the number of defects (e.g., reducing porosity (Jiang et al. 2020; Z. Wang et al. 2019) and cracking (Gao et al. 2022)) have been proposed at the research level while mature ones have been integrated to industrial solutions (e.g., PRECITEC IDM TwinTec,¹ Prima Power Laserdyne – power modulation²). In addition, predicting or even controlling the quality features of a joint such as its mechanical strength or microstructure at the weld zone has also been researched lately in numerous studies (e.g., using response surface methodology for correlating process parameters with the tensile strength value (Yaakob et al. 2019)).

1 <https://www.precitec.com/laser-welding/products/process-monitoring/precitec-idm/>

2 <https://www.primapowerlaserdyne.com/minimizing-porosity-in-fiber-laser-welding/>

With that being said, starting from the “basics”, according to the American Welding Society³, Quality Assurance includes all these actions which provide adequate confidence that a weld will perform according to design requirements or intended use. On the other hand, Quality Control is the partial or complete implementation of a Quality Assurance program where the examination of the physical characteristics of the weld and its comparison with predetermined requirements from applicable codes, specifications, other standards, and drawings is made (American Welding Society 2015).

Depending on the welding application different Quality Assurance programs exist, sharing however common steps for achieving Quality Control. Among them, Process Control and Inspection are the ones having a great impact on final product quality. Focusing on Inspection, this step is performed according to well-established procedures (standards) by certified personnel and includes operations that are performed before, during, and/or after the process. Among them, a Destructive Inspection is carried out offline while Non-Destructive Inspection depending on the method/technique can be either performed in-line or offline.

Typically, in highly automated industrial scenarios, to keep the interference between inspection and production to a minimum, for off-line testing, samples are collected during production according to specific sampling plans which ensure the statistical significance of the measured results depending on the production characteristics (e.g., batch size) and inspected off-line (Montgomery 2020).

Offline sample-based Quality Assessment

Destructive Inspection includes tests that are aiming to define the chemical, mechanical, and metallurgical properties of a specimen as well as for identifying defects that are located within the weld bead such as internal cracks, porous, incomplete fusion & poor penetration, slag inclusion, as well as internal geometrical features. These tests are conducted in the industry for qualifying a welding process, a welder, or a welding operator, and in the context of Quality Control programs and standards by organizations such as the American Welding Society and the European Welding Federation,⁴ and others.

Destructive Inspection is implemented offline, and while it tends to be replaced by non-destructive methods for several industrial applications, it still

3 <https://www.aws.org/home>

4 <https://www.ewf.be/>

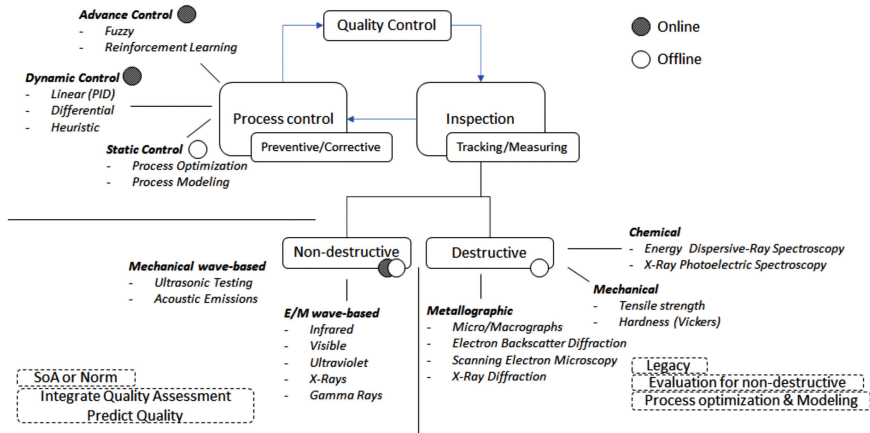


Figure 13.2 The main methods for inspection and process control of Laser Welding applications

provides the means for determining the best parameters for optimizing the process (Zhao et al. 2012), evaluating and tuning new process variants (Wang et al. 2019; AlShaer, Li, and Mistry 2014) and Non-Destructive Inspection methods and tools (Zhifen Zhang et al. 2021). For the majority of Destructive Inspections, Statistical Process Control is incorporated by using control charts for detecting and tracking abnormalities on product batches and to help identify nonconforming welding trends so appropriate attention can be applied to process design and thus achieve stability. Destructive Inspection is carried out offline either before the process (on the parts intended for welding or after the process). The most common methods/techniques have been classified into three categories for Laser Welding as described in the following subsections (Figure 13.2).

Metallographic Testing

Sectioning, mounting, planar grinding, polishing, and etching of the weld bead are standard steps for applying several inspection methods that aim to determine the geometrical features and microstructure of the weld bead profile and identify internal defects. Optical macrography (below x10 magnification) is probably the most used one for determining the geometric features of the weld bead as well as for determining internal defects such as porosity, cracks, and inclusions (Leo et al. 2015). Optical microscopy on the other hand by offering higher magnification helps resolve the grain size, the resulting solidification structure, and to some degree different phases (Popescu, Delval,

and Leparoux 2017). Among electron crystallography methods, Electron Backscatter Diffraction is most commonly used as it can rapidly characterize the spatial distribution of grains, their size, their orientation, and different crystal phases (not multiple, not similar) (Cui et al. 2012). Scanning Electron Microscopy is used also for identifying micro-geometrical features, especially on fractured surfaces resulting from mechanical testing (Casalino, Mortello, and Campanelli 2015). As regards X-ray crystallography, X-Ray Diffraction is applied when the distinction between similar crystal phases is required, and their spatial resolution is not of interest (Oliveira et al. 2022).

Mechanical Testing

Similar to the mechanical tests applied to base metals for determining their properties such as tensile and yield strength, hardness, ductility, and others, the same is also applied for determining these properties for a weld. However, they are performed in such a way that they demand preparation steps for the stress points to develop at the weld zone. For Laser Welding in particular the microstructure of the fusion and heat-affected zone is changing as regards both the size and direction of the grains and the crystal structure, compared to the base metal (Figure 13.3) (Cui et al. 2012). These changes are more profound when welding in keyhole mode, as the evaporation of alloy elements affects the hardness of the weld bead and is responsible for the generation of defects (Miyagi et al. 2018). The selection of the process parameters and the involved

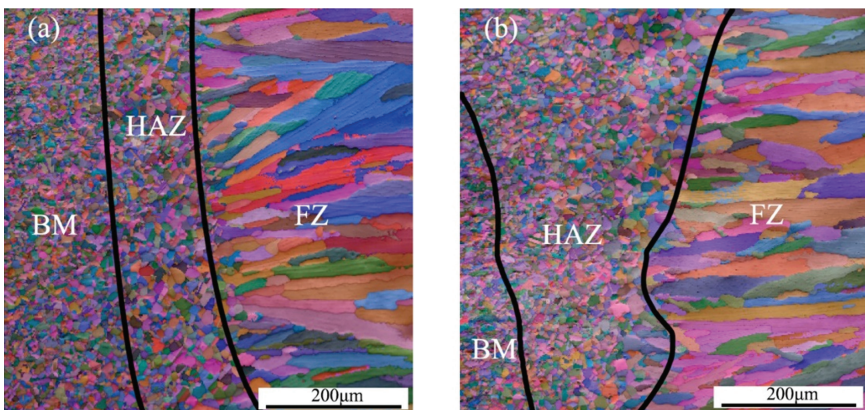


Figure 13.3 The micrographs of welded joints, including Base Metal, Heat Affected Zone, and Fusion Zone, under different welding speeds: (a) 1.5 m/min and (b) 2.5 m/min, by (Xu et al. 2020), licensed under CC BY 4.0

materials are typically determining these properties. For applications which are involving dissimilar metals, the formation of intermetallic compound layers at the fusion increases brittleness and electrical resistance (e.g., not desirable in battery welding applications) and in most cases is a point where cracks can initiate (Dimatteo, Ascari, and Fortunato 2019).

Mechanical testing is well established for welding and Laser Welding in particular as a number of testing procedures are precisely described by a number of similar standards for various joint designs and materials such as ISO 15614-11:2002 and ISO 22826:2005. Tensile shear and Vickers Hardness tests are among the most preferred ones from the research community mainly for evaluation purposes, while in the industry this may change depending on the application (Berretta et al. 2007). Finally for measuring the effects of mechanical testing beyond direct measurements of the applied forces and of the resulting global displacements Digital Image Processing is typically utilized during mechanical testing of welded samples for identifying local strain fields (Marques, Pereira, and Silva 2022).

Chemical Testing

Chemical testing is conducted on a specimen to determine its chemical composition or its corrosion resistance. In Laser Welding, the vaporization rate of specific alloy elements is greater than others. This results in a small loss of specific alloy elements, altering by this way the composition of the fusion zone. Beyond that, the use or not of shielding gas promotes the formation of crystalline solids such as nitrides or oxides affecting the purity of the joint. Surface oxides on the other hand are vaporized and dissolve into the melt-pool creating porosity. To perform simple elemental analysis of the fusion zone, Energy-dispersive X-ray spectroscopy is typically used (Nguyen et al. 2020) while to perform also compound analysis (e.g., identifying oxides, nitrides) X-ray Photoelectric Spectroscopy is most commonly used (Lai et al. 2016).

In-process, Real-time Quality Assessment

Non-destructive testing, examination, or inspection are general terms for describing the methods which are permitting the evaluation of defects without requiring the destruction of the part or specimen undergoing examination or compromise by any means of its intended functionality.

In general, Non-Destructive Inspection aims to detect imperfections both internally and at the surface of a weldment based on the interaction of a

physical quantity with the imperfection or a weld bead features. This quantity of mechanical, thermal, chemical, or electromagnetic nature is typically captured by utilizing a probing medium, or otherwise a sensor. A transduction process takes place where this quantity is converted typically to an electrical signal which represents the actual measurement and has a finite level of resolution [34]. Thus, changes in the physical quantity are expressed as changes in the output electrical signal. However, the changes in the output signals for the majority of cases cannot be directly (easily) correlated with the imperfection or weld feature, and here is where the rest of the Non-Destructive Inspection method and therefore techniques come into play.

ISO 9712:2021 makes a very clear distinction between the term method and technique as regards the Non-Destructive Inspection, e.g., Ultrasonic testing is defined as the method and when implemented, using an ultrasound probe scanning a plane, is called the C-Scan ultrasound technique.

Based on the above, in the following paragraphs, the most widely used Non-Destructive Inspection methods and techniques used in Laser Welding industrial and research applications are identified in the context of three main categories: Vision, Radiographic, Ultrasonic & Acoustic Emissions. These are incorporating both online and offline versions, with the offline ones to be widely adopted in most industrial sectors and the online ones to be mainly the subject of research studies always however with exception (e.g., vision-based inspection). Online inspection is meant to be implemented in three different stages before the process, in-process, or after the process, always however within the context of the same production line and on every part that is going to be welded, is currently welded, or has been welded. In the following subsection, the offline variant of Non-Destructive Inspection is given, but the main focus is given on the online, in-process approaches.

Vision-Based Inspection

It's probably the most widespread form of Non-Destructive Inspection and the most inexpensive one from an equipment standpoint on average. Its most basic version is conducted by naked-eye examination of the sample, while for specific inspection or monitoring applications magnification tools, and optical cameras are also incorporated. It's mainly utilized to examine the surface of the welds, during or after the process, or the part that is going to be welded (pre-inspection). For naked-eye examination measuring devices such as calipers and gauges are used. Depending on the probing medium's capabilities (resolution, accuracy, magnification, spectral response), external environmental factors (e.g., lighting conditions), and material factors (e.g., reflectivity)

different features can be identified. For example, a naked-eye visual inspection can easily reveal overlaps, undercuts, big dimensional inaccuracies, misalignments, and irregularities such as big cracks, pits, and excessive spatter in a post-process inspection (American Welding Society 2015).

On the other hand, systems already existing in the market for online vision-based monitoring and inspection such as WELDWATCHER®,⁵ MONITECH,⁶ XIRIS Automation,⁷ and others can capture process emissions (from a wide spectrum: ultraviolet, visible, and infrared) or inspect a joint right after it's done. This is done by correlating these emissions or optical signals and/or images with the weld's geometrical features or with defects in an automated manner. While the exact mechanism upon which the aforementioned systems are built is not available, a hypothesis can be made by looking at relevant research studies. Herein physics-based modeling and Data-driven approaches have been utilized for achieving such results extending however what is possible as regards predicting and identifying weld features and defects.

Starting, one of the most widely adopted vision-based setups for online monitoring is based on the integration of a photodiode or a pyrometer (coaxially or off-axis) for measuring the intensity of the visible or infrared light reflected or produced during the process (You, Gao, and Katayama 2014). Pyrometers due to their simplicity (single “pixel” detectors – generating 1D time-series signal) and low cost even as early as 2000's have been validated as per their ability to identify defects (Bertrand, Smurov, and Grevey 2000) while newer studies have augmented new capabilities by enabling the identification for example of the weld mode (conduction, keyhole, and the correlation of the measured signal with the geometrical characteristics of the weld bead) (Xiao et al. 2020). In the same vein, photodiodes have been utilized for capturing the emitted radiation which can be used for determining the weld penetration (Kaplan, Norman, and Eriksson 2009). Thus, such a mature technology has already been integrated into commercial solutions such as the LWM 4.0 system from PRECITEC⁸ and processobserver from nLIGHT plasmio.⁹

Moving on with the camera-based monitoring in Tenner et al. (2015) the authors using two high-speed cameras were able to measure the geometrical characteristics of the plasma plume and the keyhole, correlate their dynamics, and evaluate models which can predict their geometric evolution for the

5 <https://4d-gmbh.de/?lang=en>

6 <http://www.monitech.co.kr/eng/product/monitoring01.asp>

7 <https://www.xiris.com/>

8 <https://www.precitec.com/laser-welding/products/process-monitoring/laser-welding-monitor/>

9 <http://www.plasmio.eu/en/processobserver/>

different laser power settings. Similarly, in Fabbro et al. (2005) for evaluating a numerical keyhole model and an analytical model which relate the laser parameters (incident power, spot diameter) with the processing parameters (material, sheet thickness, and welding velocity), the authors used a coaxially mounted high-speed camera to measure dynamically the geometrical features of the keyhole aperture. In a novel application of Digital Image Correlation (Agarwal et al. 2018) the authors by using auxiliary illumination and creating random speckle patterns at the top surface of the workpiece were able to perform dynamic strain measurements near the fusion zone. These along with a thermomechanical model enabled the assessment of local strains and associated stress conditions to promote solidification cracking. A near-identical approach is also presented in (Hagenlocher et al. 2018) identifying characteristic strain signatures which are related to the formation of centerline cracks.

Continuing, for online inspection lately, Machine Learning and Deep-Learning approaches have been proposed that not only can perform inspection but also incorporate decision-making in the context of Quality Assurance. As such in Stavropoulos et al. (2020) the authors were able to access both the quality of part-specific parts as welds as well as the entire seam and the quality of the entire assembly (multiple seams) for a laser welding application. This was implemented through a three-stage approach where images from a Mid-Wave Infrared and a Near Infrared camera were used for extracting both statistical (Principal Components Analysis) and geometrical features for training and feeding a Support Vector Machine model which in turn feeds a Hidden Markov Model. A multi-modal sensor system of very high spatiotemporal resolution (Spectrometer, Photodiodes, Vision sensors) is used in Zhang et al. (2019) training a Convolutional Neural Network able to classify five different types of defects with above 93% accuracy. In Stavropoulos, Bikas, et al. (2022) the authors by using a vision setup were able to classify the electrical and mechanical quality of laser welded joints made for connecting battery cells by utilizing a Supervised Machine Learning schema (Figure 13.4). The list goes on and not only as regards welding applications with the majority of these Data-driven techniques being based on supervised Machine Learning approaches (Stavropoulos, Sabatakakis, et al. 2022).

Radiographic Inspection

Radiographic X-ray or Gamma-ray techniques use either film or digital detectors for the examination of welds (American Welding Society 2015). In its most simple form, radiographic testing includes a radiation source of X or Gamma rays placed in front of a specimen, while on the opposite side a

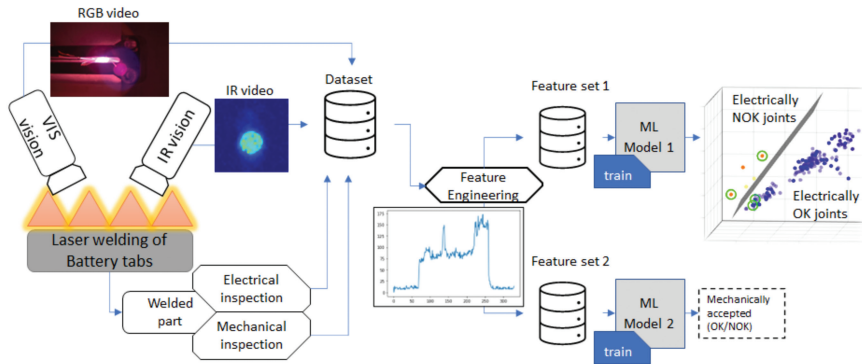


Figure 13.4 A supervised Machine Learning pipeline for incorporating quality assessment for the laser welding of battery cells (Stavropoulos, Bikas, et al. 2022)

detector medium captures the transmitted photos. An image is formed as a different number of photons reach the detector in different locations as some of them are absorbed or scattered by the weld specimens' discontinuities.

In the industry Film Radiography, computed radiography, and Digital Radiography are techniques which are producing 2D images with the last one being able to be utilized in real time (30 frames per second) (Shevchik et al. 2020). Computer tomography is another advanced radiographic technique that by utilizing complex reconstruction algorithms for combining 2D radiographic images can reproduce a 3D representation of the inspected weld's area (Dinda et al. 2016). Radiographic inspection is mainly used for identifying internal discontinuities such as cracks, lack of fusion, incomplete penetration, corrosion, and fit-up conditions while depending on the technique and equipment feature differences can be measured. In the industry, Radiographic Inspection is used and interpreted mainly for detecting porosity by qualified personnel offline usually with no samples or specimens (American Welding Society 2015; Chauveau 2018). While some Radiographic Inspection equipment is fast enough to be utilized online, there are not any related products, at least to the author's knowledge. This is maybe due to the fact that the Radiographic Inspection equipment requires a lot of space as well as physical access on both sides of a weld.

On the other hand, in the research community, computer-aided Radiographic Inspection is an active research field as human interpretation has various limitations (difficulty to recognize small flaws within a short time, often short of objectivity, consistency & intelligence, labor-intensive). In general, the majority of these studies focus on the automatic detection of defects (segmentation and/or classification) from radiographic images/

films however without any particular focus on laser welding application. Early ones are based mostly on heuristic algorithms or classic Machine Learning models, using feature descriptors (Valavanis and Kosmopoulos 2010; Vilar, Zapata, and Ruiz 2009) while the latest ones are based on Deep Learning models such as Convolutional Neural Networks for classification (Yang et al. 2021). An extensive review of this field can be found in Hou et al. (2020).

Online approaches have also been implemented especially (Figure 13.5) for process monitoring (Heider et al. 2013; Boley et al. 2019; Miyagi et al. 2017; Hummel et al. 2021; Wasmer et al. 2018) while also integrating quality assessment with the use of Deep Learning models (Shevchik et al. 2020). Finally, as with the visual examination, Radiographic Inspection is used as the basis (providing the ground truth) for developing other inspection approaches (de Bono et al. 2017) as well as for process parameter optimization (Cheng et al. 2021; Farrokhi, Larsen, and Kristiansen 2017) and parameters for laser application and welding in general.

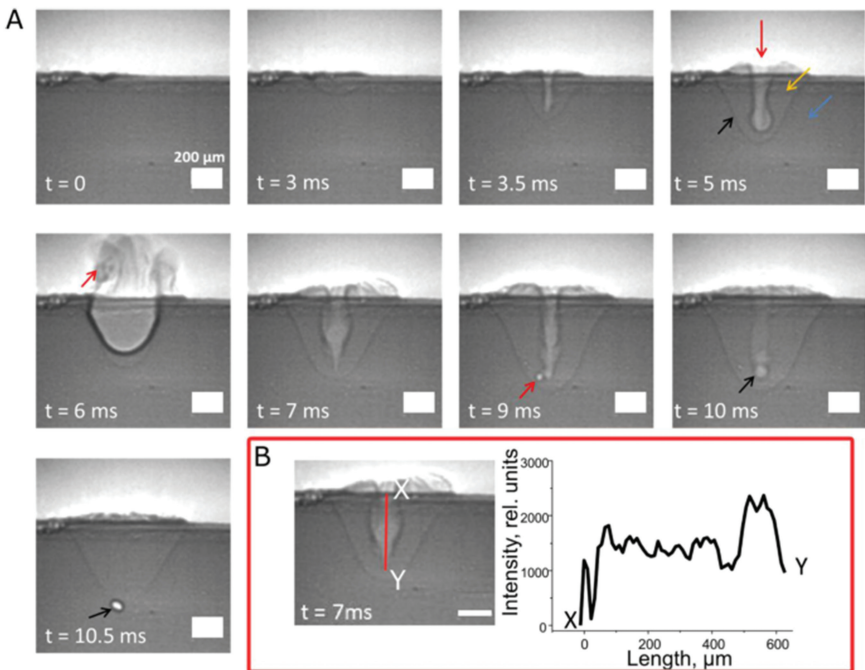


Figure 13.5 In-process X-ray snapshots of the aluminum sample exposed to a single laser pulse, Supervised deep learning for real-time quality monitoring of laser welding with X-ray radiographic guidance, by Shevchik et al. 2020, licensed under CC BY 4.0

Ultrasonic Inspection & Acoustic Emissions

This type of inspection can be divided into two subcategories namely ultrasonic testing and acoustic emissions testing. Starting with Ultrasonic Inspection this type of examination is based on the direction of an ultrasonic wave into the specimen across a predictable path and measuring its reflection across different points (American Welding Society 2015). Ultrasonic Inspection is used for detecting and evaluating internal defects; however, the thickness of the material is a limiting factor. The sensitivity of Ultrasonic Inspection increases with the frequency used. Conventional Ultrasonic Inspection is performed in two modes, “contact mode” and “immersion mode” using typically piezoelectric transducers for generating and receiving ultrasonic waves or more rarely using Electro Magnetic Acoustic Transducers and Capacitive Micromachined Ultrasonic Transducers while other advanced Ultrasonic Inspection includes Phased Array Ultrasonic Testings and Laser Ultrasonics (Honarvar and Varvani-Farahani 2020). For detecting macro-defects, guided waves are used and for detecting microscale defects bulk waves are utilized. However, most commonly Ultrasonic Inspection is done using bulk waves through longitudinal waves (Olisa, Khan, and Starr 2021). Ultrasonic Inspection can be performed in various modes (A/B/C/D/F & P Scan) and welding is used for detecting internal defects such as porosity, cracks, voids, inclusions incomplete fusion, and others.

In the industry Ultrasonic Inspection is typically performed offline by qualified personnel; however, there are some products dedicated to the inspection of girth welds in the pipe industry that can do the automatic interpretation. With that being said, in the research community, Ultrasonic Inspection testing has been incorporated in various ways for improving the testing procedure in general. One branch of this research tree concerns the development of Artificial Intelligent-based tools for supporting or automating the interpretation of the Ultrasonic Inspection results in post-process offline inspection, e.g., defect identification (Sambath, Nagaraj, and Selvakumar 2011; Martín et al. 2014), or correlation of these results with attributes of the weld (e.g., tensile strength (X. Wang et al. 2019), nugget diameter (Ghafarallahi, Farrahi, and Amiri 2021)) that otherwise will be available through destructive testing. The other branch includes a small number of research that are proposing online in-process monitoring or quality assessment approaches based on heuristic or Artificial Intelligent-based algorithms. For example, in Mizota, Nagashima, and Obana (2015) the authors were able to monitor the melt pool depth for a Gas Tungsten Arc Welding of a thick steel plate using the Phased Array Ultrasonic Testing technique, while in another case the authors by integrating an ultrasonic transducer into the electrode retrieved several ultrasonic signals

eigenvalues in the frequency and time domain and used them to develop a quality evaluation application that can predict the nugget's diameter and thickness with a low error (less than 7%) and the overall quality with an accuracy (binary) of 96% (Hua et al. 2019).

Acoustic-Emissions inspection is another type of inspection which is based on the utilization of mechanical waves. Unlike Ultrasonic Inspection, Acoustic-Emissions are performed online and, in the process, require the acoustic emissions generated during the welding process to be captured (Stavridis, Papacharalampopoulos, and Stavropoulos 2018). Typically involves a microphone (contactless measuring) or a resonance sensor placed near the region of interest (contact measurement) and up until now, there are very few commercial solutions for the welding industry.¹⁰ In the same pattern as the rest of the Non-Destructive Inspection methods, Acoustic-Emissions inspection has been incorporated to analyze the mechanism generation of Acoustic-Emissions signals and correlate them with process features such as the plasma plume and the penetration of weld bead for the case of pulsed laser welding (Luo et al. 2019). Data-driven approaches have also been undertaken for this reason where for example Neural Networks and multiple regression analysis were used to find that Acoustic-Emissions time-frequency features during welding of steel such as Sound Pressure Deviation and Band Power and process parameters were linearly correlated with the weld penetration (Huang and Kovacevic 2011). Quality assessment has also been integrated using Acoustic-Emissions extracted features and the Supervised Adversarial Sequence Tagging method for classifying weld quality into five states: good, excessive penetration, burn-through, porosity, and porosity-excessive penetration achieving a prediction accuracy of 91.18% (Asif et al. 2020).

Beyond the above-mentioned Non-Destructive Inspection methods which are the most widespread at the industrial level, it is worth also to mention some less-known ones which have been identified according to ISO 9712:2021 Magnetic Particle Examination, Panetran Examination, Eddy current examination, Leak Testing, Ferrite content Examination, and Infrared Thermography which as with the previous subsection mainly concern in-process and post-process inspection. On the other hand, pre-process inspection for laser welding is mainly performed by optical means for providing feedback for seam tracking and identifying non-acceptable fitting conditions. One of the most well-established methods for seam tracking which has been also incorporated for the post-inspection of the seam profiles is based on the identification of the distorted laser line profile typically from a monochrome optical camera which is typically formed at the interface of the workpieces that are going to be

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welded (Luo and Chen 2005). Lately, another method called Magneto-Optical Imaging has also been used for seam tracking tasks as it enables the detection of micro gaps as small as 0.1 mm (X. Gao, Liu, and You 2014; Gao et al. 2017). However, as with the previous technique, Magneto-Optical Imaging has also been used for post-process inspection for the identification of cracks (Gao et al. 2020). This method utilizes the magnetic field distribution as configured from a workpiece placed within a magnetic field by introducing a special filter that changes the polarization of light passing through it and thus visualizes in a sense the direction of the magnetic field which depends on the continuation of the involved material.

As this section was meant as an introductory to Non-Destructive Inspection describing the general concept, and advancements and not a review article, the reader is advised to refer to the following studies for a more detailed analysis (American Welding Society 2015; Chauveau 2018; Marques, Pereira, and Silva 2022).

Process Modeling and Control

Up until this point, a brief review of inspection for laser welding was made analyzing in a sense these actions that enable the measuring of the key performance indicators for a joint and the assessment of its quality. However, quality assurance includes beyond inspection also process control to ensure that a weld will be made according to its design standards. Laser Welding in its baseline form is not deterministic, meaning that for a given weld design and a set of process parameters, reproducibility in terms of quality is not guaranteed as several external factors can act as disturbances to the process. However, for every new design, initial process optimization is always required in order to fall within the specification boundaries and minimize defects. In the industry, a New Production Introduction is typically accompanied by trial-and-error sessions in which the various process parameters are tweaked in order to achieve that or to improve production and product quality in general. This procedure required time and resources, and while methods have been developed for narrowing down the problem of studying all the design factors and the number of required experimentations, they are simply just an initial step (Xiansheng et al. 2011).

Static Control – Optimization and Modeling

To this end, process optimization tools have been available on the market in the form of simulation software, while in the research community novel approaches have been proposed both for multiscale and dynamic simulation

of Laser welding and for extracting the best optimal process parameters statically. Starting from the last one, typically a Response Surface Optimization method is applied for a set of process parameters and a set of response variables (quality indicators) as they've emerged from the factorial experiment process. The resulting surfaces are used for selecting the optimal process parameters (Zhao et al. 2012). Moving to the Simulation approaches, while their initial configuration of the corresponding models is time-consuming, they can be parameterized to simulate more than one welding scenario offering greater flexibility and accuracy and process insights that cannot be easily obtained unless complex monitoring setups and extensive destructive testing are carried out. Indicatively in Ke et al. (2021), the authors developed a numerical model to investigate the mechanisms of bubble formation during aluminum laser welding and how beam oscillation can help select the appropriate motion patterns for suppressing porosity. In Halim et al. (2021) on the other hand, a three-dimensional transient model has been developed for the Laser Welding of dissimilar materials and enabled the numerical prediction of the thickness and location of the intermetallic compounds. In the same vein, a plethora of studies have been conducted over the years with the ultimate goal to cover a wider spectrum of material and dynamic phenomena, especially during key-hole welding (Pastras et al. 2015).

Proportional Integral Derivative (PIDs) and Differential Models

Process modeling and optimization methods, in general, can indeed account for meeting the design requirements however cannot address the dynamic variations of the welding environment or variations in the operation of the welding machinery. In the research community, this has been tackled by PID-based control approaches and on a smaller scale by regulators that rely on differential models of the actual process.

As regards the first type, the most common process parameter selected for control is the laser power (Konuk et al. 2011; Božič, Kos, and Jezeršek 2020), while a vision type of sensor provides the required feedback which is compared to a reference value for determining the power input for the subsequent step. Before that, the relation of the feedback signal with the quality feature has been established (e.g., measured temperature from a thermal camera vs penetration depth) either by using a simple regression model or Deep Learning. The second major type of control is based on differential models which describe different process phases (e.g., heating, melting, solidification) for adaptive control. However, the tuning of these models is not trivial since many factors

are involved in the previously mentioned phases (Papacharalampopoulos, Stavropoulos, and Stavridis 2018).

Beyond these two main types of closed-loop control, other approaches are also existing. The trajectory of the laser beam is another process parameter that is subjected to real-time control based on the signal of the seam tracking sensor for achieving increased dimensional accuracy (Regaard, Kaierle, and Poprawe 2009).

Advanced Control – Data Driven

While the previously mentioned control approaches have shown some potential, they are still experiencing some serious challenges. As such, the ones based on PIDs are trying to linearize the non-linear welding dynamics while requiring time-consuming tuning in order to be applied even to similar cases. On the other hand, differential, model-based control requires deep knowledge of the underlying physical mechanisms.

To address these challenges recent studies have experimented with the adoption of Reinforcement Learning schemas (Günther et al. 2016; Masinelli et al. 2020). This means in general that the controller (agent as it is called in Reinforcement Learning) interacts with its environment to shape its control strategy. In more detail, an interpreter translates signals from capture from the environment and passes them to the agent, in turn, the agent makes a decision based on the input signals and acts on its environment. The result of this action is identified with the help of the interpreter, and based on a reward policy the agent receives a score for his actions adjusts his control strategy and continues from the beginning (Figure 13.6).

Digital Twins and Cyber-Physical Systems

A Digital Twin is the digital representation of a physical object, process, and/or environment of which its states are changing based on a set of inputs/output values that are in complete sync and constant cross-validation with the corresponding ones from its real analogy. A Digital twin, while it tries in a sense to emulate the states of its physical analogy, it is also able to detect abnormalities and trends in its current or feature states, and thus prompts or apply corrective and/or optimization actions (Stavropoulos and Mourtzis 2022).

The application of Digital Twins in manufacturing has been foreseen in different ways, from their utilization as virtual test benches (e.g., process

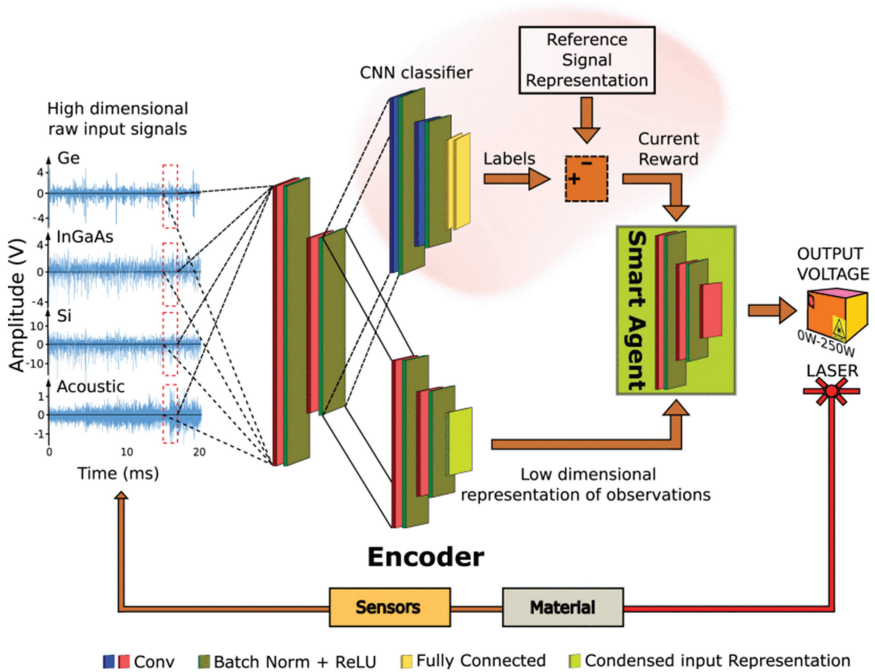


Figure 13.6 Reinforcement learning schema as described for achieving Adaptive Control for Laser Welding, Structure of the complete control unit made up of three main building blocks: an encoder that processed the data from the sensory input to retain only the quality critic events, a smart agent interacting with the welding process, and a feedback network based on a convolutional neural network for quality monitoring, by Masinelli et al. 2020, licensed under CC BY 4.0

optimization, test process capabilities against design requirements) and/or data manipulation (e.g., Model generated data, Data aggregation & correlation from multiple sources) to process monitoring & control (e.g., develop & validate data-driven and physics-based models, predictive maintenance and process status, closed-loop control) (Stavropoulos 2022).

Looking at the previous approaches for advance inspection and process control of Laser welding, it's clear that they are tackling quality assurance from a very limited scope, which is something that can be said also for several other welding and manufacturing processes in general. The operation of Digital Twins inherently requires feedback and connection to different levels of a production system in order to represent and act as accurately as possible to the target entity. This implicitly enables “root cause analysis” and eventually Quality Control generalization which contradicts the current approaches

for Quality Assurance that address specific use cases and key performance indicators (Papacharalampopoulos et al. 2020).

The successful integration of a Digital Twin is strictly related to the quality of the data that exchanges with the physical space. A straightforward connection with a network of sensors and actuators cannot just simply accomplish that. To this end, the Cyber-Physical System is another concept that seems to provide a framework for setting an environment in which a Digital Twin can be hosted and act. In contrast to the Digital Twins, a Cyber-Physical System focuses on providing a design framework for interconnecting a system's physical assets and computational capabilities (Stavropoulos 2022). In the following paragraphs, a brief review of the most representative one is made both for process-oriented Digital Twin and Cyber Physical System.

Digital Twin

Among the most representative frameworks of the Digital Twin concept that can be utilized for implementing Digital Twins toward quality assurance of laser welding applications has been described in Söderberg et al. (2017) where the defect prevention measures start as early as the concept phase and do not focus only on specific points in the production (e.g., during the welding process). In the same vein and moving one step further the authors in Franciosa et al. (2020) are proposing a Digital Twin framework that is validated in the context of a New Production Introduction scenario for the remote laser welding of aluminum car doors. The Digital Twin is engaged for all three levels which characterize the different development phases of the product, i.e., concept, scale-up, and production pilot. By fusing a set of tools for data acquisition & management, statistical analysis & Machine Learning, and Computer Aided Engineering Simulation, the Digital Twin maps the probability of different requirements and parameters to occur for each development phase and ultimately enables optimization of the involved parameters. In more detail for the concept phase, a preliminary optimization of the part-to-part gap with respect to the welding process parameters and design of the clamp layout was implemented. In the scale-up phase, the refinement of the final clamp layout was implemented by utilizing multiscale data (microscopy, 3D scanning, temperature & mechanical testing data) while during welding in a real-time closed-loop control model handled any process variation concluding for a reduced commissioning time and overall right-first-time (no cracks, discontinuities, lack of gap bridging) above 96%.

In a more recent study (Zhehao Zhang et al. 2022), the authors developed a process-exclusive Digital Twin for seam tracking predictive control in laser

welding. The main components of the Digital Twin are a virtual representation of the welding environment, which is constructed based on globally assembled optical images, and the system model, which handles the prediction of the future seam position and the generation of the optimal control input. The welding environment enables the extractions of seam positions which feed a Gaussian process model which in turn predicts the subsequent seam positions and enables predictive control. Finally, the approach is validated for three different seam types presenting improved robustness which outperforms the traditional PID controller ones.

Beyond “complete” Digital Twin approaches, tools and methodologies which can be integrated or aid in the development of Digital Twin for Laser Welding applications have also been proposed lately. Indicatively in Papacharalampopoulos, Sabatakakis, and Stavropoulos (2021), the authors are investigating how different physic-based modeling methods can be fused to form a knowledge base library using their best attributes as regards the simulation of the physical phenomena taking place during welding (heat transfer, phase change).

At this point, the concept of Digital Twin is an active research field, and apart from a few specific attempts which are evaluated through laser welding applications the majority of related studies could be applied for that too considering that they are evaluated on other welding or thermal-based processes in general. For example, in Stavropoulos et al. (2021) the adoption of the Digital Twin framework enables uncertainty management for robust process control of laser additive manufacturing processes, while in Wang, Jiao, and Zhang (2020) the authors developed a deep learning-empowered digital twin which replicates joint growth and uses this information as feedback for penetration control and evaluated it in the context of gas tungsten arc welding.

The main challenges related to the adoption and industrialization of digital twins for Laser Welding and Manufacturing, in general, are not related to a single factor according to (Lattanzi et al. 2021). Herein one of the challenges for deploying Digital Twins vision suffers from the synchronization between the physical and the digital world. Data and interface/protocol heterogeneity hinder or even make impossible the uniform data handling and processing and thus the construction of semantic data models. Another challenge concerns the fidelity of the involved models which is consolidated with the model’s level of detail. While an over-simplified model may limit the performance of the Digital Twin, a high-detailed one may result in additional requirements which are adding complexity both from a computational and a hardware point of view (e.g., a high number of sensors). In addition, adapting the virtual models’ errors between virtual and physical entities is another active research topic that

hasn't converged just yet. Finally, as regards the model-related issues, another challenge concerns the "knowledge" of the corresponding model about faults on the real assets which not only requires identifying a fault but also "simulating it" in advance, which acts as a serious limiting factor for fault prevention. Moving on to the data management aspects of the Digital Twin, the fusion of data from multiple sources (real-time data from a sensor, data from a simulation, etc.) which are related to an asset (e.g., with a weld on a part assembly) comes with several challenges. Structuring and associating these data to make exploitable from subsequent models is not trivial.

Cyber-Physical System

In most cases, Digital Twins implicitly are developed in the context of a Cyber-Physical System without this however it is always applied in reverse. For Laser Welding, the Cyber-Physical System concept has been adopted as provides the designing elements for structuring quality assurance systems that can natively integrate data-driven approaches for quality assessment and control (Stavropoulos et al. 2020). However, studies that explicitly investigate the utilization of the Cyber-Physical System for Laser and welding quality assurance applications, in general, are fairly limited (Ahmed et al. 2021; Stavropoulos et al. 2020).

Again, as with the Digital Twin, Cyber-Physical System approaches for other welding applications or similar manufacturing processes could be considered to be applicable also for laser welding. For example, in Ahmed et al. (2021) the Cyber-Physical System concept is applied for Resistance Spot Welding to collect, store, and utilize the weld test data for supporting weldment design and predicting weldability and certification. This was done by adopting a generic custom architecture of different layers, modules, and processes architecture which offers the required abstraction for being applied to other welding applications and separates its operational features from the particular application. On the other hand, Cyber-Physical systems as already mentioned have been conceptualized as the information sources for Digital Twin for integration planning (Biesinger et al. 2019).

Conclusion

Quality Assurance for welding applications has evolved both with the utilization of new technologies and methods as well as with the adoption of new

concepts. For Laser Welding, this is apparent both at an industrial and research level where several products or studies have already been developed and target Right-first-time and Zero-Defect strategies. And why may indeed present exceptional performance their scope is limited, meaning that they are applicable at best for a fairly limited number of cases. The same goes for their reconfiguration/readaptation which cannot be addressed without a significant amount of time or resources and of course without human supervision. To this end, the concept of Digital Twins seems to tackle quality assurance holistically and not drain its diagnostic capacity to specific points during production until it clarifies which is the root cause. At the same time tools like Reinforcement Learning and new concepts such as Active Learning (van Houtum and Vlasea 2021) seem to have the potential for automating the reconfiguration of quality monitor and control tools, which although presenting great performance as mentioned earlier are not autonomous nor mature enough to support new trends in production such as massive customization. In Active Learning, the learning algorithm queries the user to label specific data points that are critical for the performance of the corresponding quality assessment model thus it greatly limits the need for vast amounts of labeled data, something that is typically required from classic Supervised Learning approaches for them reach high performance.

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EVA SOLEDADE VAZ MARQUES
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