

**DYNAMIC STRAIN AGEING BEHAVIOUR OF 316L AUSTENITIC
STAINLESS STEEL**

**A THESIS SUBMITTED TO
THE GRADUATE SCHOOL OF NATURAL AND APPLIED SCIENCES OF
KARABUK UNIVERSITY**

BY

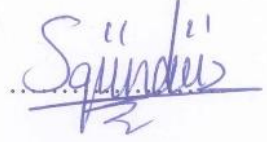
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**IN PARTIAL FULFILLMENT OF THE REQUIREMENTS FOR
THE DEGREE OF DOCTOR OF PHILOSOPHY IN
DEPARTMENT OF
MANUFACTURING ENGINEERING**

December 2018

I certify that in my opinion the thesis submitted by Guma Alnaji MOHAMED titled "DYNAMIC STRAIN AGEING BEHAVIOUR OF 316L AUSTENITIC STAINLESS STEEL" is fully adequate in scope and in quality as a thesis for the degree of Doctor of philosophy.

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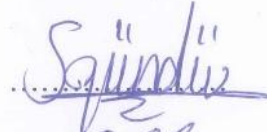
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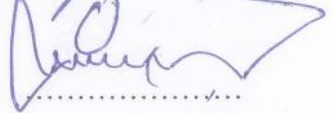
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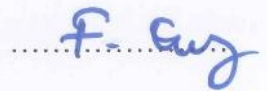
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“I declare that all the information within this thesis has been gathered and presented in accordance with academic regulations and ethical principles and I have according to the requirements of these regulations and principles cited all those which do not originate in this work as well.”

Guma Alnaji MOHAMED

ABSTRACT

Ph.D Thesis

DYNAMIC STRAIN AGEING BEHAVIOUR OF 316L AUSTENITIC STAINLESS STEEL

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In this thesis, dynamic strain ageing phenomena in AISI 316L austenitic stainless steel was investigated under as-received, as-welded, and weld metal conditions. Tensile test was carried out on as-received, as-welded, and weld metal conditions for the temperatures of 25, 100, 200, 300, 400, 500, 600, 700, 800°C at a strain rate of $1 \times 10^{-3} \text{ s}^{-1}$. Microstructure and fracture surfaces were investigated by optic and scanning electron microscopes (SEM). 316L austenitic stainless steel showed different dynamic strain ageing behavior under as-received, as-welded, and weld metal conditions, which are discussed in terms of microstructure and mechanical properties.

Keywords : Stainless steel, dynamic strain ageing, welding, weld metal.

Science Code : 902.2.042

ÖZET

Doktora Tezi

316L ÖSTENİTİK PASLANMAZ ÇELİKLERİN DİNAMİK GERİNİM YAŞLANMA DAVRANIŞLARI

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Bu çalışmada, 316L östenitik paslanmaz çeliğinin dinamik deformasyon yaşlanma davranışı ana malzeme, kaynaklı bağlantı ve kaynak metali şartları için araştırılmıştır. Ana malzeme, kaynaklı bağlantı ve kaynak metalinden elde edilen çekme test numuneleri 25, 100, 200, 300, 400, 500, 600, 700, 800°C test sıcaklıklarında ve $1 \times 10^{-3} \text{ s}^{-1}$ gerinim oranında çekme testine tabi tutulmuştur. Farklı şartlarda ve sıcaklıklarda test edilen numunelerin mikroyapı ve kırık yüzey incelemeleri optik ve tarama elektronmikroskopunda (SEM) yapılmıştır. 316L östenitik paslanmaz çelik ana malzeme, kaynaklı bağlantı ve kaynak metali şartlarında farklı dinamik deformasyon yaşlanma davranışı göstermiştir. Bu durum mikroyapı ve mekanik özellikler göz önünde bulundurularak tartışılmıştır.

Anahtar Kelimeler : Paslanmaz çelik, dinamik yaşlanma sertleşmesi, kaynak, kaynak metali.

Bilim Kodu : 902.2.042

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LIST OF NOMENCLATURES

ASS	Austenitic Stainless Steel
BCC	: Body Centered Cubic
FCC	: Face Centered Cubic
SA	: Strain Ageing
SSA	: Static Strain Ageing
DSA	: Dynamic Strain Ageing
YS	: Yield Strength
UTS	: Ultimate Tensile Strength
PLC	: Portevin – Lechatelier
RT	: Room Temperature
SEM	: Scanning Electron Microscope
SFE	: Stacking Fault Energy
EDS	: Energy Dispersive Analyser
HAZ	: Heat Affected Zone

SYMPOLS AND ABBREVIATIONS INDEX

LOG	:	Logarithmic
α	:	Ferrite
γ	:	Austenite
σ	:	True Stress
ε	:	True Strain
K	:	The Strength Coefficient
n	:	The Work Hardening Index
Mn	:	Manganese
N	:	Nitrogen
C	:	Carbon
Cr	:	Cromium
Ti	:	Titanium
FN	:	Ferrite Number

CHAPTER 1

INTRODUCTION

In the last days the main reason for the increased use of stainless steels that are among the indispensable materials of today's industry is their high corrosion resistances which they show without losing their mechanical properties in corrosive environment. Stainless steels that can be found in every type and shape and easily shaped, can be sourced successfully to further application fields with improved resource methods. Stainless steels must contain 12% chromium in their combinations in order to have stainless properties; chrome makes the steel small grain, by reducing the critical cooling speed have a severe hardening effect and increase the amount of oxidation resistance in high temperatures [1].

The presence of more than 12% chromium in the steel as the alloying element protects the steel against oxidative acids such as nitric acid as well as from adverse effects of atmosphere, although only chromium containing steels are not resistant to acids such as hydrochloric acid and sulfuric acid. These acids lift the surface of the chromium oxide layer that protects the surface, thus the steel remains unprotected. In today's industry, stainless steel with alloy elements such as nickel, molybdenum are produced which have good resistance to reducing acids. These alloying elements can change microstructures of the steel effectively, which also helps to classify stainless steels [2].

Austenitic stainless steels are the most important types of stainless steels which are known as nonmagnetic, formable and weldable. They are used from cryogenic temperatures to the hot temperatures of furnaces and jet engines. They contain chromium in the weight percentage of 16 and 25, and they can also contain nitrogen in solid solution, both of which contribute to their high corrosion resistance

[1]. 316 type austenitic stainless steels contain molibdenium which increases corrosion resistance and strength at high temperatures. Mechanical properties are similar to that of 304 type stainless steel, but this alloy shows higher strength at higher temperatures [2].

In the early 1990s, intergranular cracking was observed in 316 Nuclear Grade (NG) and 304L stainless steel materials. In several reported cases, cracking was observed in the heat-affected zone (HAZ) and as-received metal of the weld joint. Although all parameters affecting the cracking are so far unknown, deformation could be an important parameter [3]. The thermally dramatic act of casting a small amount of metal as in each weld run can severely alter the microstructure and toughness of the parent steel in the HAZ, locally changing the original tough parent steel into a less tough and sometimes brittle material. The thermal contraction accompanying freezing may also produce short and/or long range stresses in the surrounding materials. Then stresses may be sufficient to cause local plastic strain around the weld bead [4].

The affecting mechanisms for intergranular cracking may also be dynamic strain ageing (DSA), which occurs at elevated temperatures depending on strain rates. The aging occurs slowly at room temperature and faster at higher temperatures, because diffusion of the solute atoms responsible for the ageing is supported by increasing the temperature. When the ageing occurs in alloys containing solute atoms, which can rapidly move to dislocations and prevent their movement during deformation, the phenomenon is commonly known as dynamic strain ageing [5].

AISI 316 and 316L austenitic stainless steels show dynamic strain ageing behaviour in the temperature range of 200–800°C [6-8]. However, literature investigation indicated that there are inconclusive works related to dynamic strain ageing behaviour of weld joints of austenitic stainless steels. Therefore, an effort is needed to study the effect of dynamic strain ageing on the mechanical properties of austenitic stainless steels which have been used at elevated temperatures. The purpose of the present study is to investigate dynamic strain ageing behavior of

316L austenitic stainless steel under as-received, as-welded and weld metal conditions. This investigation provided the metallurgical basis to understand the mechanical behaviour of as-received, as-welded and weld metal samples of 316L austenitic stainless steel dynamically aged at elevated temperatures.



CHAPTER 2

GENERAL INFORMATION ABOUT STEELS

2.1. INTRODUCTION

Steels which are alloys of iron and other elements, especially carbon, are widely used in construction and other applications due to their low cost and high tensile strength. Carbon, other elements and inclusions in iron function as hardening agents that prevent the dislocations that occur in the crystal lattices of the iron atoms. Figure 2.1 shows the steel classification chart [3].

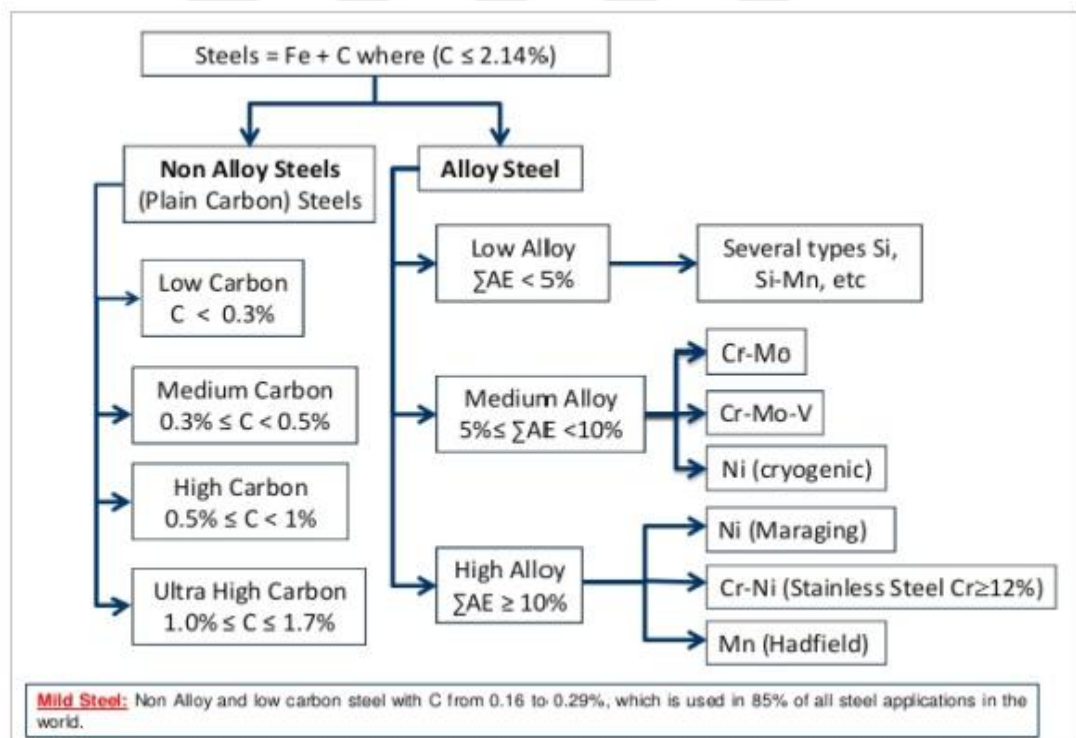


Figure 2.1. Steel classification chart [3].

The carbon in typical steel alloys can contribute up to 2.1% of its weight. The change in the amount of alloying elements, their formation in steel, either dissolved phases or precipitated phases, delay the movement of these dislocations which make the iron relatively strong and thus control properties such as ductility, tensile strength and hardness of the resulting steel. The strength of steel compared to pure iron is only possible at the cost of ductility, which is an excess of iron [9].

Although steel has been produced in bloomer ovens for thousands of years, the use of the steel industry has expanded since the 17th century, when more efficient production methods have been developed. A massive new steel era began in the mid-19th century with the invention of the Bessemer process. This was followed by Simiemen-Martin process and Gilchrist-Thomas that improved the quality, respectively. With introductions, light steel modified wrought iron. Figure 2.2 shows Fe-Fe₃C diagrams in which cast iron pieces and steels can be seen [10].

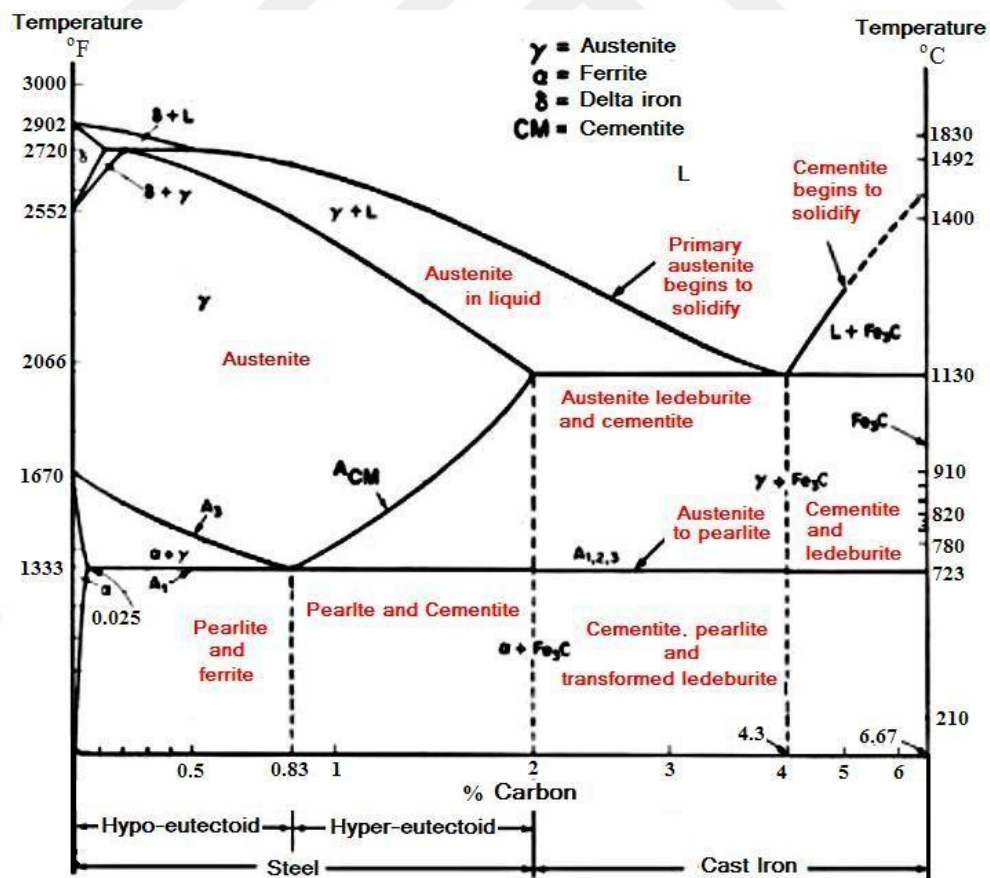


Figure 2.2. Fe-Fe₃C phase diagrams [10].

Further improvements in processes such as basic oxygen steel production have taken the older methods to a greater extent, further increasing the quality of the metal and reducing the cost of production. Today, steel is one of the most common materials in the world, producing up to 1.6 billion tons a year. Steels are an important component in infrastructure, buildings, ships, vehicles, machines, automobiles, weapons and tools. The American Iron and Steel Institute (AISI), which is based on asserted standards organizations, has broadly categorized steel into three groups according to their chemical composition:

- Carbon Steels
- Alloy Steels
- Tool Steels

2.1.1. Carbon Steels

Carbon steels, which make up 90% of total steel production, contain trace elements of alloying elements and can be further categorized into 3 groups according to the carbon contents

- Low carbon steels/mild steels contain up to 0.3% carbon
- Medium carbon steels contain 0.3 – 0.6% carbon
- High carbon steels contain more than 0.6% carbon [10].

2.1.2. Alloy Steels

Alloy steels contain alloying elements (eg manganese, nickel, silicon, titanium, chromium, copper and aluminum) at various ratios to manipulate the properties of the steel such as corrosion resistance, hardenability, strength, weldability formability or ductility. Automobile parts, pipelines, transformers, electric motors and power generators are applications for steel alloys [10].

2.1.3. Tool Steels

Depending on the grade or quality of the steel, it is a type of metal alloy containing low levels of iron and mostly carbon. Alloy steel, however, is any steel in which one or more elements of the carbons are intentionally added to achieve more desirable properties. Manganese, silicon, vanadium, molybdenum, nickel and chromium are the most common elements that are added to produce alloyed steels [10].

The difference is somewhat uniform, but in order to make it distinguishable, all steels that are alloyed with more than 8% of the weight of carbon and alloying elements are considered high alloy steels. Alloy steels are harder, more resistant and durable to corrosion. It is difficult to weld all steels with medium and high carbon levels. However, if the carbon levels are reduced to 0.1-0.3%, such alloyed metals can achieve better strength and thus have greater weldability and formability [10]

2.2. STAINLESS STEELS

2.2.1. General Properties of Stainless Steel

The steels are oxidized in the atmosphere like most of the other iron alloys and an oxide layer (rust) is formed on the surfaces. This oxide layer, which covers the surface of the steel opposite to the protective oxide layer that is formed on the surface of aluminum and zinc, does not prevent the oxidation to proceed to the inner parts. In stainless steels, corrosion resistance is increased depending on the amount of chromium being increased. Although there are many different views on this issue, the most widely accepted is that a tight and thin chromium oxide layer is formed on the steel and this layer prevents the progression of oxidation and corrosion. The protective effect of the chromium in the steel content comes from a finite balance between chromium and oxygen. When the chrome-containing steels are not covered with a chromium oxide layer, they are very susceptible to corrosion and especially oxidation. This situation is active, however, when this oxide layer is found the possibility of formation, protects the metal against corrosive environments, so the steel becomes passive. The limit and degree of passivation depends on the activity of

environment and on the type and composition of stainless steel. If the conditions are appropriate, passivation is permanent and stainless steel has a very slow corrosion rate [11].

When this passive film disappears and there are no conditions to reproduce, stainless steel may be exposed to corrosion. In this respect, for the formation of corrosion resistance, stainless steel should contain at least 12% chromium and there is need of the presence of oxygen in the environment [11]. In order to form a passive layer on the surface of the stainless steel, there is no need for a chemical process. This film suddenly occurs when the surface is in contact with oxygen, so the passivation process is based on the principle of removal of free iron, oxides and other surface impurities on the surface [11].

Stainless steels contain not only elements that make up the stainless steel feature, but also alloying elements added to meet some other requirements or inevitable carbon and some impurities [12]. Carbon is a strong austenite builder. It is added to high strength alloys for hardening and strength additive effect. It affects the corrosion resistance of weld metal and toughness at low temperature in the negative direction.

Chromium is a carbide and ferrite builder. It is an alloy element that provides corrosion and scaling resistance. It has no significant effect on creep and strength resistance at high temperatures in stainless steels. Nickel is a strong austenite builder and stabilizer. It is added to prevent grain growth at high temperatures in high chromium and low carbon steels. It increases strength and affects the toughness of the weld metal under zero temperature in the negative direction. Manganese is an austenite maker and increases the cracking resistance of weld metal in full austenitic alloys.

Aluminum is a strong ferrite builder. It increases resistance to scaling at high temperature. It reduces the effect of aging hardening by incorporating some high strength alloys with titanium. It is a strong nitride builder. By joining to the weld metal containing 12% chromium, makes the building ferritic, so hardener. Niobium is a powerful carbide builder. It is used to stabilize austenitic stainless steels against

sedimentation of chromium carbide. It is a medium moderate carbide builder. It is added to high strength alloys to influence hardness and strength. It is added to some martensitic stainless steel grades by binding carbon to reduce the steel hardening trend. Nitrogen is a strong austenite builder. It is added to prevent grain growth at high temperatures in high chromium and low carbon steels. It increases strength and affects the toughness of the weld metal under zero temperate in the negative direction.

One of the sulfur, phosphorus and selenium elements is added to molybdenum or zirconium in small amount in order to increase the susceptibility of the machining of stainless steel. These three elements encourage cracking in the weld metal. Silicon is a ferrite builder. It is used to increase corrosion resistance in austenitic steels. It increases scaling resistance at high temperature. It is added to enhance the carburization resistance of steels that will be used at high temperatures. Titanium is used as a balancing element to prevent sedimentation of chromium carbide in austenitic stainless steels. It is a strong ferrite builder. It is added to some high temperature resistant alloys due to the effects of hardness and strengthening. It is added together with aluminum to some high resistant and heat resistant alloys to affect aging hardening. Tungsten is a strong ferrite builder. It is added to increase the creep resistance and the strength of some high temperature alloys.

The iron-chromium system forms the basis of the group of stainless steels. Chromium stabilizes ferrite which has body centered cubic crystal lattice (BCC) structure. Closes the austenite field, that is formed by austenite with a face centered cubic crystal lattice (FCC) structure in the iron-chromium equilibrium phase diagram and this phase shows the highest chromium solubility (12%) at about 1000°C. As can be seen from the iron-chromium equilibrium phase diagram (Figure 2.3), no austenite is found at temperatures above 1390°C and below 830°C in any region [13].

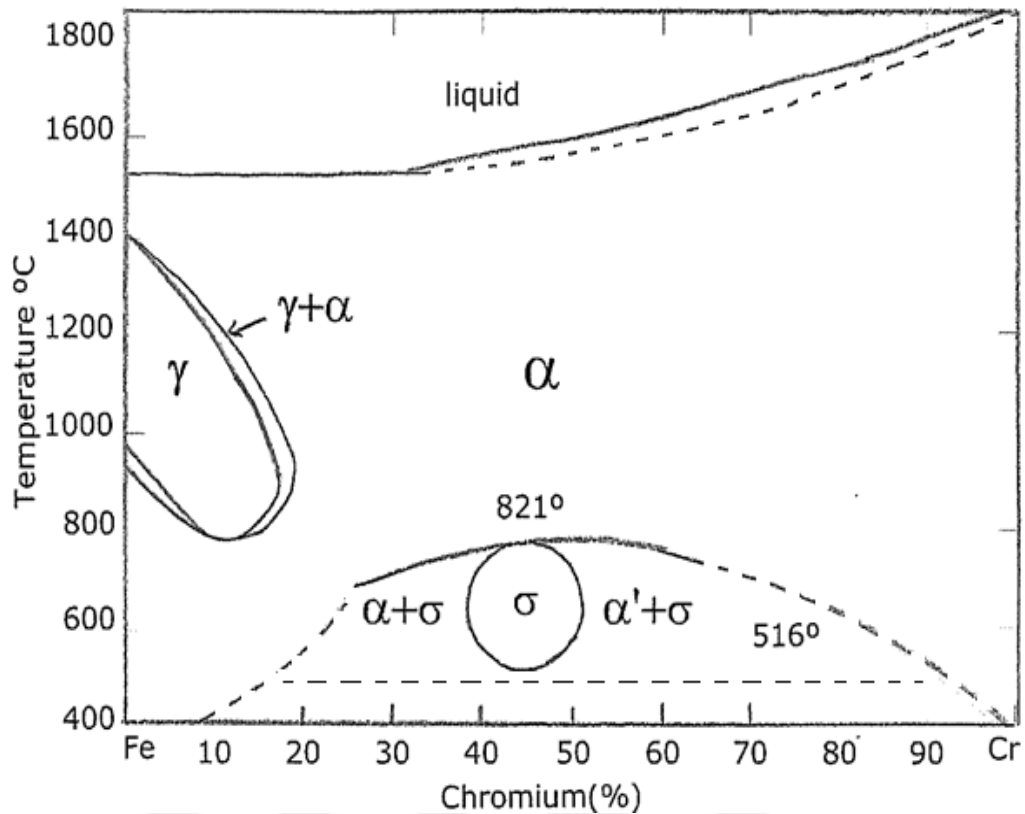


Figure 2.3. Iron-chromium equilibrium phase diagram [13].

Pure iron-chromium alloys, as shown in the binary equilibrium diagram in Figure 2.3, in case of more than 12% chromium content, when they solidify from liquid state, they solidify as BCC crystal lattice ferrite and also shows no transformation during cooling. This event is caused by the ability of the chromium to form ferrite which sharply also limits the stable austenite field.

The area in a pocket shape located on the upper left of the diagram γ called austenite region, expands depending on the amount of carbon in the content of the alloy. Another element is nitrogen that expands the γ field. The carbon and nitrogen, which are very strong austenite stabilizers, increase the solubility of chromium in the austenite by expanding the γ field. For example, if the amount of carbon + nitrogen accesses to 0.13%, $\alpha \rightarrow \gamma \rightarrow \alpha$ conversion is not observed, the chromium content should be increased to 24%, if the amount of carbon + nitrogen increases to 0.2%, the amount of chromium should be increased to 27%.

The σ -phase seen in the lower middle region of the iron-chromium equilibrium phase diagram is a tetragonal cage. This phase occurs at a very slow cooling rate of 800 - 600°C. From time to time, in the various commercial stainless steel types, this σ - phase, which is seen a result of a hard and brittle metal compound by exposing to specific temperatures for a long time, the occurrence range is too narrow the iron-chromium binary system. At temperatures below 520°C this phase is divided into phases $\alpha + \alpha'$ by an eutectoid transformation. This solid-state reaction has a very slow rate of formation and development. Fast cooling prevents the formation of σ - phase and in this case α' phase occurs by decomposing below 520°C. The α' phase contains approximately 90% chromium, is a brittle phase and the result of this precipitation is encountered as an event called 475°C relaxation.

There is only a 0.2% difference between the lattice parameters of the $\alpha + \alpha'$ phases, but in spite of this, α is iron rich, α' is chromium poor. The iron-nickel equilibrium phase diagram used today is generally accepted as valid for temperatures above 450°C (Fig 2.4). Solubility of nickel in α iron reaches the maximum value between 400 - 500°C. Nickel is a violent austenite stabilizer. γ area of 30% nickel content expands from 500°C to 1450°C [14]. Therefore, the M_s temperature shows a great decrease with the increase in nickel content. For example, 20% nickel 220°C, 30% nickel 0°C, 34% nickel -220°C.

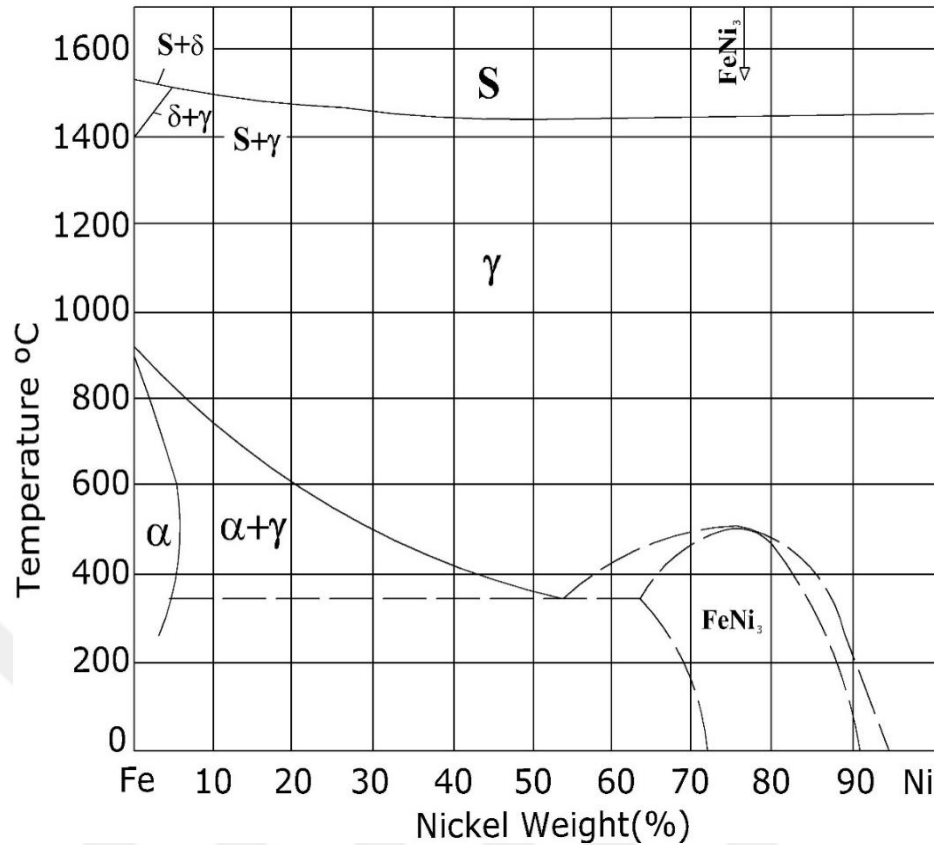


Figure 2.4. Iron-nickel equilibrium phase diagram [14].

The events below 450°C has not yet clarified in the iron-nickel equilibrium phase diagram. As can be seen from the chromium-nickel equilibrium phase diagram, pure chromium possesses an allotropic transformation (Figure 2.4).

2.3. TYPES OF STAINLESS STEEL

2.3.1. Ferritic Chromium Stainless Steels

Ferritic chrome stainless steels have a fairly wide range of uses in industry due to their being one of the first stainless steel types to be developed following their discovery. They contain 16-30% chromium, depending on the content of the alloying elements, especially the amount of carbon, magnetism, can be cold rolled or hot rolled, but the toughness, ductility and corrosion resistance show maximum value when normalized. Their machinability and corrosion resistance are superior to those of martensitic chromium stainless steels [7]. Ferritic chrome stainless steels are more

economical than chrome-nickel austenitic stainless steels due to their nickel-free content, which is an expensive and strategic element and this is a very important advantage in today's conditions.

The other advantages can be listed compared to the austenitic chrome-nickel steels as follows:

- More resistant to stress corrosion cracking in chlorinated solvents.
- Higher yield strength.
- Show less severe cold strain hardening.
- Capable of magnetism.

These types of stainless steels, which have a very bright and decorative appearance, are second in the global stainless steel market with their largest market share immediately after austenitic stainless steels. Ferritic chrome stainless steels, being economic and the superior features mentioned above, has a wide range of uses such as in internal and external architecture, steam production and transmission equipment, automotive industry, appliance construction, kitchen and household appliances, chemical and petrochemical industry, food industry, boiling water containers and pipes.

The coefficient of thermal expansion of such steels is close to that of low-alloy steels. This feature avoids the problems that would arise due to thermal expansion difference if they are used together with plain carbon steels in large constructions as in architectural constructions. However, the coefficient of thermal conductivity of such steels is half that of normal steels.

Ferritic chromium stainless steels are Fe-Cr-C alloys that contain sufficient amounts of chromium to provide corrosion resistance in the composition or chromium with ferrite stabilizing elements such as aluminum, niobium, titanium and molybdenum to prevent the formation of austenite at high temperatures. According to the AISI standard, such steels grouped in the 4XX series as martensitic steels are represented

as X6Cr17 (430), X10CrN28 (446) as they are in the form of high alloy steels according to DIN 17440, EU88, EU95 and TS 2535 (Table 2.1) [15, 16].

TS 2535 is a nickel-free, generally ferritic stainless steel with up to 0.2% C containing 11.5 to 18% chromium in its ferritic stainless steels' composition. They are defined as “magnetized up to 750°C, they can not be hardened by heat treatment”, but now excludes ferritic stainless steels containing up to 30% chromium and up to 4% nickel in some newly developed species. Since this event was prepared by TS 2535 in 1976. Table 2.1 shows the typical ferritic stainless steels, application areas and combinations, which are also given by AISI and DIN norms [6].



Table 2.1. Main ferritic chromium stainless steels [6].

AISI (UNS)*	EURONO RM DIN (TSE)	Chemical Composition (%)								
		C	Mn	P	S	Si	Cr	Ni	Mo	Other
405(S40500)	X6CrAL13	0,0 8	1,0 0	0,0 40	0,0 30	0,5 0	11,50/1 3,00			0,10/0,30 Al
409(S40900)	X5CrTi12	0,0 8	1,0 0	0,0 40	0,0 30	1,0 0	11,50/1 3,50			6XC/0,75Ti
429(S42900)	X7Cr14	0,1 2	1,0 0	0,0 40	0,0 30	1,0 0	11,50/1 3,50	1,25/2 ,50		
430(S43000)	X6Cr17	0,1 2	1,2 5	0,0 60	0,1 50	1,0 0	12,00/1 4,00		0,60*	
430F(S43020)	X12CrMos 17	0,1 2	1,2 5	0,0 60	0,0 60	1,0 0	12,00/1 4,00			
430FSe(S43023)	-	0,1 2	1,0 0	0,0 40	0,3 00	1,0 0	12,00/1 4,00			0,15Se(min)
434(S43400)	X8CrmO17	0,1 2	1,2 5	0,0 60	0,1 50	1,0 0			0,60*	
436(S43600)	-	0,1 2	1,0 0	0,0 25	0,0 25	0,7 5	11,00/1 3,00	0,50/1 ,00	0,75/1 ,25	5X C/0,70 Nb+Ta
442(S44200)	-	0,2 0	1,0 0	0,0 40	0,0 30	1,0 0	15,00/1 7,00	1,25/2 ,50		
446(S44600)	X10CrN28	0,2 0	1,0 0	0,0 40	0,0 30	1,0 0	16,00/1 8,00		0,75	0,25N
*Can be joined by the manufacturer										

Commonly used types of such stainless steels are X6CrAl13 (405), X6Cr17 (430), X12CrMoS17 (430F) steels. In the X6CrAl13 stainless steel, 0.1-0.3% of the aluminum content is balanced with the ferritic structure to prevent the formation of martensite in spite of the low chromium composition. X6Cr17 steel contains 14 - 18% chromium, the change in chromium content gives the user a choice between various properties. As the amount of chromium decreases, better weldability, hardness, strength and toughness are obtained, but corrosion resistance is reduced. When the amount of chromium is increased, the corrosion resistance is increased but this time the mechanical properties are weakened, so the amount of chromium is limited between 16-18%. Titanium and niobium stabilized species (X6CrTi17 and X6CrNb17) of the X6Cr17 steel exhibit high corrosion resistance in the absence of highly concentrated corrosive solutions. With the addition of sulfur, the machinability of the chip is increased, enabling the use of the X12CrMoS17 ferritic stainless steel in the production of stainless steel bolts and cutting tools. The highest chromium-containing type of ferritic stainless steels is the X10CrN28 (446) core as defined by the AISI norm. This steel has very high corrosion and oxidation resistance. Especially used in construction of running machine parts in sulfur atmosphere.

New types of stainless steels have been developed containing small amount of carbon and nitrogen that can be economically produced (Table 2.2). The most important types of these steels according to the AISI norm are 444 (18Cr-2Mo) and 26-1 (26Cr-1Mo) steels. When they are stabilized with elements such as titanium and niobium, when welded, these new types of ferritic stainless steels, which are resistant to intergranular corrosion, also have good resistance and toughness against cracking of the grooves and tensile corrosion cracking which may occur in chloride environments [6].

Table 2.2. New types of ferritic stainless steels [6].

Alloys	C and N (%)	Stabilization Elements (%)
Fe-%18 Cr %1Mo-Ti	C0,025 (max) N 0,025 (max)	Ti+Nb, 0,20+4(C+N)min, 0,80 max
Fe-%25 Cr %1Mo	C0,005 (max) N 0,015 (max)	Nb, 13-29 (N)
Fe-%26 Cr %1Mo-Ti	C0,040 (max) N 0,040 (max)	Ti, 0,2-1,0
Fe-%28 Cr- %2Mo-% 4 Ni	C0,015 (max) N 0,035 (max)	Nb, 12(C+N)+0,2
Fe-%29 Cr %4Mo	C0,010 (max) N 0,020 (max) C+N 0,025 (max)	-
Fe-%29 Cr- %4Mo- %2Ni	C0,010 (max) N 0,020 (max) C+N 0,025 (max)	-

As these types of stainless steels are usually high chromium and very little carbon, sometimes no austenite or sometimes very small amount of austenite is formed during cooling from high temperatures or from liquids state. For this reason there is no austenite-ferrite transformation. However, as shown in the Fe-Cr equilibrium diagram (Figure 2.3), 17% chromium steel containing 0.1% C contains a small amount of austenite at around 1000°C. The amount of this austenite depends on the nitrogen and carbon contents, and also on the relative amounts of austenite and ferrite forming elements. The expanding effect of the carbon and nitrogen on austenite field is given in Figure 2.5 as a function of chromium content. During the heating process of the steel, passing through the edge of the γ ring, first the amount of austenite increases and then decreasing at high temperature [16].

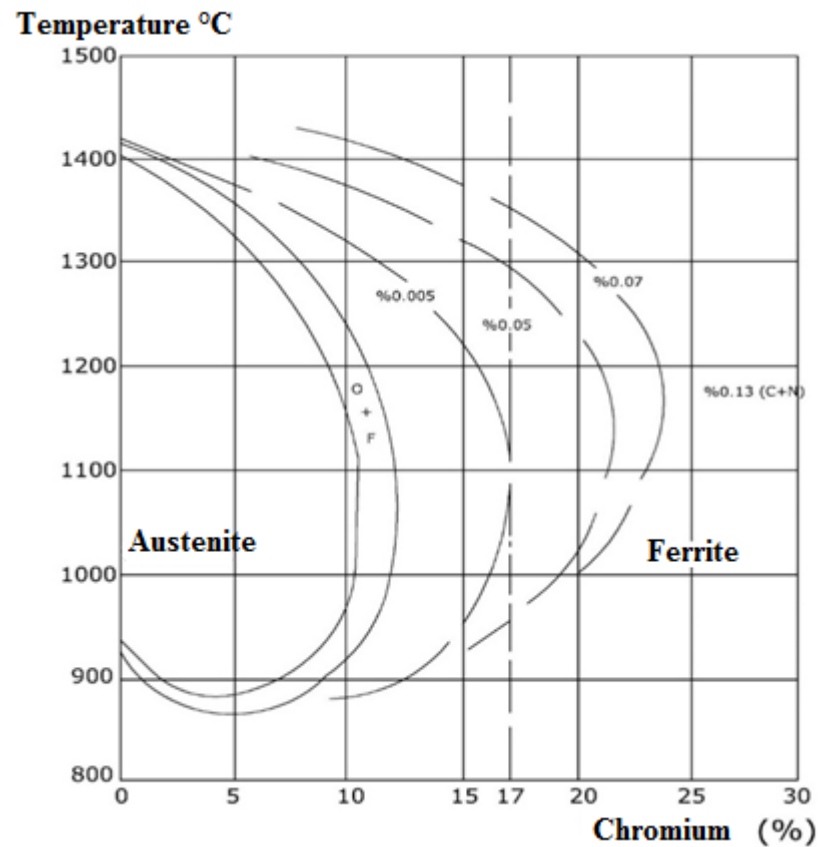


Figure 2.5. Effect of carbon and nitrogen (equal weight given in $\gamma + \alpha$) on the γ field in the Fe-Cr balance diagram [16].

The mechanical properties of ferritic chromium steels are closely related to internal structures and show very important differences with austenitic chromium-nickel steels. The values in Table 2.3 belong to the normalized material.

Table 2.3. Mechanical properties of some normalized ferritic stainless steels [6].

UNS	AISI	Product Format	Flow Strength (MPa)	Tensile Strength (MPa)	Elongation At 50 mm, %	Shrinkage %
S40500	40	Sheet	276	448	25	-
S40900	409	Sheet or Band	241	448	25	-
S43000	430	Sheet or Band	345	517	25	-
S44200	-	Rod	310	552	20	40
S44600	446	Sheet or Band	345	552	20	-
S44600	446	Rod	345	552	25	45

In ferritic stainless steels, the grain size should not be kept away from the effect of internal structure and properties. The increase in grain size reduces the toughness, but affects the strength at a slower rate. The easier realization of atomic movements in ferritic crystal systems, ferritic chromium steels exhibit faster grain coarsening than chromium-nickel austenitic steels at high temperatures, require to take into account grain blooms, which lead to reduced toughness in the production and processing of such steels. In this respect, the heat treatment and hot deformation of ferritic stainless steels must be under strict control. The highest reduction rate that can be achieved should be applied in hot forming. For example, in titanium-stabilized 17% chromium ferritic steels, a reduction of more than 60% is required to achieve grain inspection in hot rolled steel.

Moreover, since the grain sizes of ferritic chromium steels are rapidly increased with increasing holding time and temperature at which the hot process finishes is also very important. Recrystallization after cold treatment is affected by grain size, strain rate and annealing conditions. Even with large deformation rates, large grain formation can be encountered at high annealing temperatures. But the second recrystallization is result of a rapid heating. The addition of molybdenum also reduces the recrystallization temperature as it slows the diffusion at the ferrite core. As a result, grain structure is further examined and the probability of secondary recrystallization

is weak. Carbide and carbonitride deposits in stabilized ferritic steels are stable to high temperatures and thus delay grain refinement [17].

Figure 2.6 shows the changes in mechanical properties at different temperatures in ferritic chromium steels. The flow and tensile strengths rise rapidly with decreasing temperature. However, high purity steels should not be neglected to exhibit high elongation and shrinkage even below room temperature. Ferritic chromium steels have high yield strengths up to 300°C compared to austenitic chromium-nickel steels. Chromium ferritic steels exhibit lower strength properties than austenitic steels due to the fact that atomic movements are easier in ferritic steels at higher temperatures. The results of the bending-rotation test applied to the steel with 17% chromium indicate that these steels are not lower than the austenitic steels in terms of fatigue. Depending on the structural changes at the source, changes in the fatigue strength of the welded joints can be seen. Fatigue strengths were better in molybdenum-containing alloys as grain bloom slowed down. Precipitation and at the same time the grain size affects the fatigue strength. Sudden cooling of chromium steels from high temperatures also reduces fatigue strength. 1 hour annealing at 350°C and cooling in air improves fatigue properties [17].

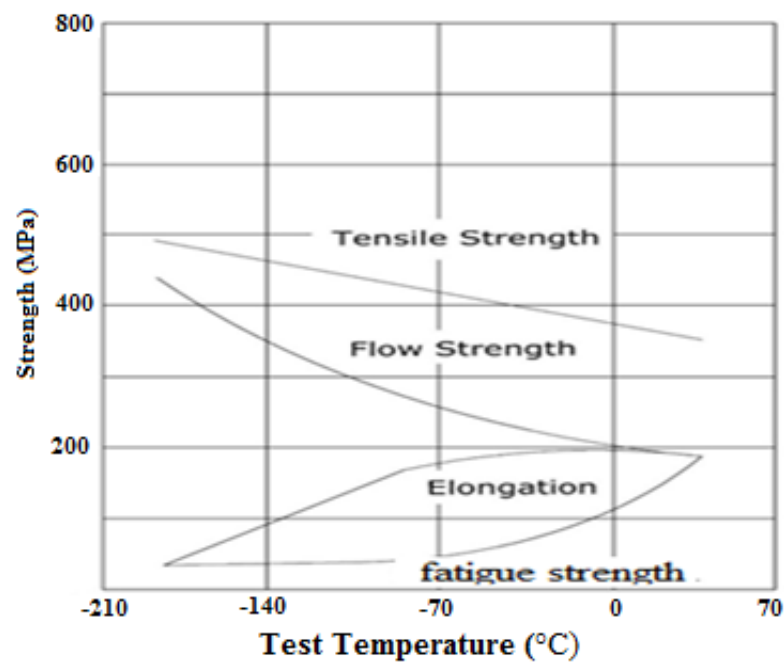


Figure 2.6. Influence of subzero temperatures on mechanical properties of 17% chrome steel [17].

As can be seen from Figure 2.6, chromium steels ferritic, due to be in terms of toughness they show a transition zone, depending on the temperature. This behavior is characteristic of BCC metals. Carbon and nitrogen precipitates have the greatest effect on the toughness of ferritic steels in relation to their origin. It is possible to reduce the transition zone to very low temperatures by reducing the carbon content. The nitrogen effect is much less in this regard. The effect of grain blooms in 0.01% nitrogen-containing alloys, assumption of that very finely dispersed precipitates in the grain are effective in reducing the disruption of dislocations, is strong. The type of embrittlement described above can stabilize the result of titanium or niobium and tantalum stabilization. The above-mentioned type of embrittlement can be stabilized by titanium or niobium and tantalum stabilization [17].

These steels are mostly produced as cold rolled sheet, wire or small diameter rods due to the fact that grain size in thick section is not properly controlled. Small carbon and nitrogen contents and their precipitates are utilized to reduce thick sections. The addition of molybdenum reduces the tendency to grain growth. Oxygen, hydrogen, sulfur and phosphorus have a great influence on the toughness of ferritic steels, so these elements should be kept as small as possible in steel [17].

2.3.2. Martensitic Chromium Stainless Steels

Martensitic chromium stainless steels are stainless steels containing more than 11.5% chromium and having an austenitic structure at high temperatures and martensitic at room temperature with a suitable cooling process. This definition limits the chromium content of stainless steels because the steel composition must fall into the γ area at high temperatures in order for this transformation to take place. Carbon, as has expanded its γ ring, in practice steel containing 18% chromium can also be converted to a full austenitic structure at high temperature, enters this group. In such stainless steels, chromium is limited to 11.5% at the bottom and 18% at the top. The lower limit is the corrosion resistance and the upper limit is the ability to transform the steel into a fully austenitic structure at high temperatures [17]. Figure 2.7 shows type of the stainless steels.

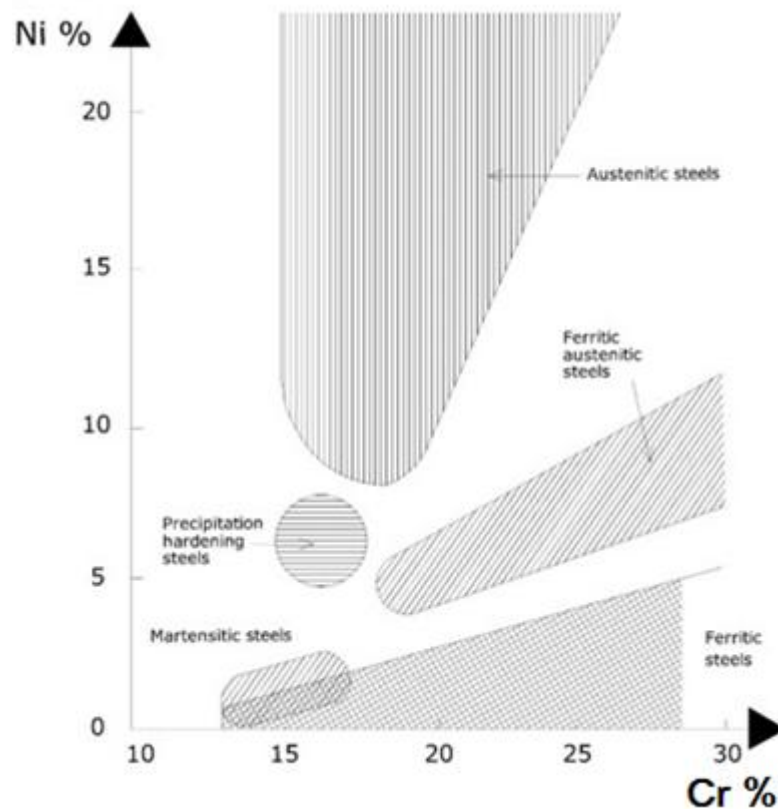


Figure 2.7. Types of stainless steel.

According to AISI norm, such steels grouped as 4XX series are symbolized as a group of high alloy steels, marked in the X10Cr13 and X105CrMo17 style according to DIN and TSE standards. While TS 2535 stainless steels are defined as 'resistant to various chemical effects and containing 11.5% by weight of very chromium in their composition', martensitic stainless steels are defined as 'martensitic structure stainless steels having a composition of 11.5-18% chromium and hardenable by heat treatment' [18]. Excluding chromium-nickel martensitic stainless steels, which are soft martensitic steels, chromium content of 440 species (16-18% chromium content) does not exceed 14% and the sum of other alloying elements is not more than 2-3%.

Martensitic stainless steels are more difficult to process than normal carbon steels. It improves the ability to process small amounts of sulfur in X12CrS13 (416) steel and selenium in 416Se steel in the AISI. The use of selenium is less effective than sulfur in terms of corrosion resistance. High carbon species such as X90CrMoV18 (440B) and X105CrMo17 (440C) have high strength, high corrosion and abrasion resistance;

So these types find application in the construction of cutting tools, valves and roller bearings [18].

Since the critical cooling rates of martensitic stainless steels are very low, martensite occurs even in slow cooling. Corrosion resistance of such steels is very good in martensitic, hardened state. They do not lose their stainless resistance up to 815°C, only if they are exposed to high temperature for a long time, slight corrosion will be started, in this regard, they can not be used continuously at temperatures above 700°C in the industry. These steels, which are hardened and tempered like high alloy steels, are magnetic and have very superior strength properties.

The tensile strength of martensitic stainless steels is much higher than that of low alloy steels and carbon steel with the same carbon content. The high strength seen in AISI 414 and 431 types is due to the low toughness of this steel containing a certain amount of nickel and therefore the austenite area has been lowered to lower temperatures and the resulting tempering temperature is not increased. Effect of tempering temperature on mechanical properties of 410 stainless steel with 12% chrome is shown in Figure 2.8. As seen, the greatest effect on the mechanical properties of such steels is the tempering temperature. The tempering to be carried out at higher temperatures causes a rapid decrease in tensile properties despite the increase of the section narrowing and elongation [17].

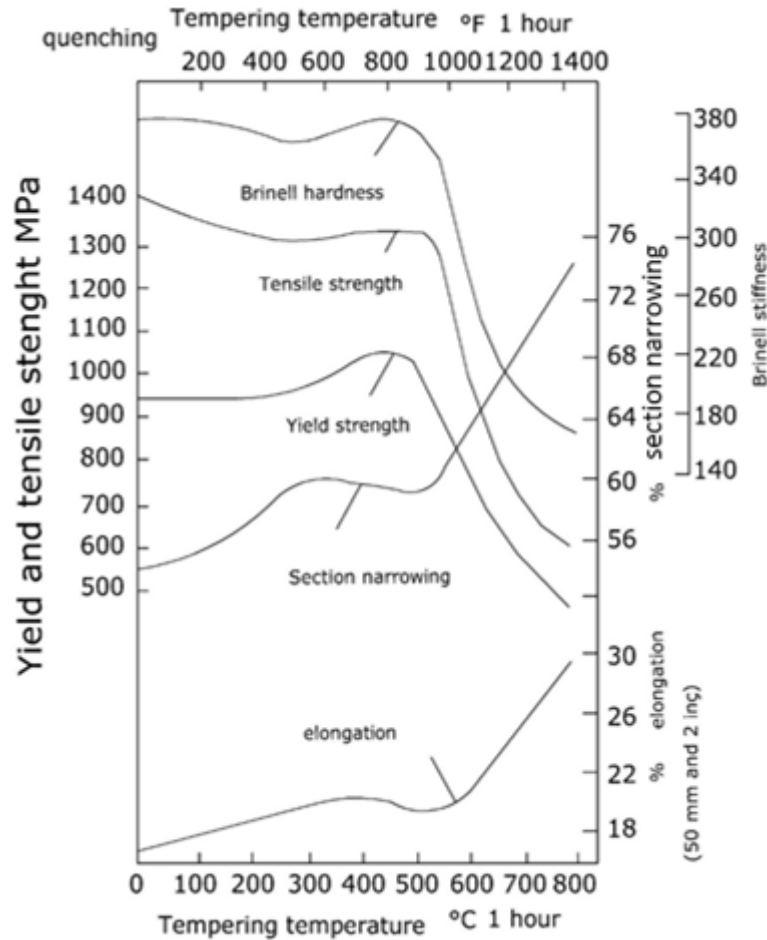


Figure 2.8. Effect of tempering temperature on mechanical properties of 410 stainless steel with 12% chrome [17].

Steels used at high temperatures must contain molybdenum, chromium and silicon elements. Here, molybdenum is added to increase strength at high temperature, chromium is added to prevent graphitization and oxidation, and silicon is also added to increase the resistance to oxidation. In this respect, more ferritic, austenitic or precipitation hardened stainless steels are used at high temperatures. However, X10Cr13 (410) steel containing 12% chromium shows corrosion and oxidation resistance up to 700°C. The mechanical properties of this steel up to 650°C are quite satisfactory. In this respect, the 410 steel has a very wide application area in the refining of steam turbine wings, petroleum refineries [17]. The notch-impact transition temperature of martensitic stainless steels is slightly below room temperature. In this regard, these steels are not used in the production of parts operating at below zero temperatures.

Martensitic stainless steels are completely austenite at about 1000°C in the limit values of the amounts of carbon and chromium they contain. From this temperature by rapid cooling produces maximum martensite in the microstructure. When heated up to temperatures between 820 and 960°C, no austenite forms completely and by cooling from this temperature range, microstructure of ferrite and martensite can be formed. In the hardened state of these steels, the toughness is low and a tempering heat treatment is usually required for proper toughness. Tempering temperature can be adjusted to provide different strength levels. The low carbon chromic-nickel martensitic stainless steels are permanently watered and tempered. Depending on the steel type, quenching is normally carried out following a heat treatment at 950-1050°C and followed by a tempering at 600°C [17].

2.3.3. Precipitated Hardened Stainless Steels

Precipitated hardened stainless steels are a family of Fe-Cr-Ni stainless steel alloys containing alloying elements such as copper, molybdenum, niobium, aluminum and titanium which exhibit precipitation hardening by the effect of one or more of these elements. Precipitation hardening is, in principle, an aging process following rapid cooling applied after the precipitation of the alloy. The alloy elements in the steel mentioned above dissolve during the solution annealing process and they precipitate as submicroscopic particles during the aging process to increase the hardness and strength of the matrix. As a result of this process, steel can have the mechanical properties of martensitic stainless steels and the corrosion resistance of AISI 304 (X5CrNi1810) austenitic stainless steel. The most important advantage that such steels have in production is the improvement of their mechanical properties by applying a heat treatment at 480-600°C after they are easily processed and formed in their normalized state. Their strength can be up to about 1700 MPa, so that values above the strength of martensitic stainless steels can be achieved [17].

Although the principles of precipitation hardening of stainless steels were not known in the 1930's, research on such stainless steels has continued in World War II, and after 1946, the production of the first commercial precipitation hardened stainless steel, which was called "stainless W", has been started. After this year many new

precipitated hardened stainless steels have been developed to find application areas in aircraft, space and defense industries. Today's precipitated hardened stainless steels are widely used on the outer surfaces of high speed aircraft, missile bodies, marine vehicles, fuel tanks, aircraft landing gears, pumps, nuts, bolts, cutting tools and concepts. Precipitated hardened stainless steels are classified into three types depending on the structural change and properties of the resultant steel due to the post-annealing processes. These are;

- Martensitic precipitation hardened stainless steels
- Semi-austenitic precipitation hardened stainless steels
- Austenitic precipitation hardened stainless steels [18].

The precipitation hardened stainless steels and their chemical compositions, known in commercial practice as non-standard qualities, are included in the 6XX series in the AISI standards but can not be found in the technical literature in this series. This grouping and structural differences between precipitation hardened stainless steels directly influence the weldability and thermal behavior of such steels. Hence, it is useful to examine the metallurgical characteristics of such steels. While martensitic precipitation hardened stainless steels are completely austenitic during the solution annealing at temperature range of 1050-1080°C, the austenite transforms into martensite at cooling and has a completely martensitic internal structure. The martensitic conversion starting temperature (M_s) is 95-150°C. Then, these steels are aged by heating to a temperature range of 480-650°C.

One or some of the elements such as molybdenum, titanium, copper, niobium and aluminum in the steel stock may be very small as a result of this heat treatment, the hardness and the strength are increased. Table 2.4 shows chemical composition of the martensitic precipitation hardened stainless steels applied to different heat treatment and used in the industry. The mechanical properties of the steel vary depending on the aging temperature and the temperature process, so that it can meet the specific requirements of use. As a result of the applied heat treatments, the martensitic precipitation hardened stainless steels are separated into two groups depending on the change in tensile strength. For example, types such as 17-4 PH, 15-

5 PH, Custom 450, and stainless W have moderate strength with tensile strengths up to 1378 MP, PH 13-8 Mo and custom 455 with tensile strengths higher than 1378 MPa [19].

Table 2.4. Chemical composition of the martensitic precipitation hardened stainless steels applied to different heat treatment.

Symbol	Chemical Composition									
	Cr	Ni	C	Mn	Si	Cu	Mo	Ti	Al	Other
Martensitic Precipitation Hardened Species										
Stainless Wb	16.75	6.75	0.07	0.50	0.50	-	-	0.80	0.20	-
17-4 PH	16.50	4.25	0.04	0.40	0.50	3.60	-	-	-	Nb+Ta 0.25
15- 5PH (XM-12)	15.00	4.60	0.04	0.25	0.40	3.50	-	-	-	Nb+Ta 0.35
CROLOY 15-5PH	15.75	7.50	0.03	0.80	0.45	-	-	0.60	0.40	-
CUSTOM 450 (XM-25)	14.90	6.50	0.03	0.30	0.25	1.50	-	-	-	Nb+Ta 0.75
CUSTOM 455(XM-16)	11.75	8.50	0.03	0.20	0.20	2.25	-	1.20	-	Nb+Ta 0.30
PH 13-8 Mo (XM-13)	13.00	8.00	0.04	0.05	0.05	-	-	-	1.00	-
ALMAR 62 (XM-9)	14.50	6.50	0.03	0.30	0.20	-	-	0.80	-	-
IN 736	10.00	10.00	0.02	0.10	0.10	-	2.00	0.20	0.30	-
Semi-Austenitic Precipitation Hardened Species										
17-7PH	17.00	7.00	0.07	0.70	0.40	-	-	-	1.15	-
PH 15-7 Mo	15.00	7.00	0.07	0.70	0.40	-	2.25	-	1.15	-
AM-350	16.50	4.25	0.10	0.75	0.35	-	2.75	-	-	N 0.10
AM-355	15.50	4.25	0.13	0.85	0.35	-	2.75	-	-	N 0.12
PH 14-8 MoC (XM-24)	15.50	8.75	0.05	0.10	0.10	-	2.50	-	1.35	-
Austenitic Strain Hardened Species										
17-10 P	17.0	10.50	0.12	0.75	0.50	-	-	-	-	P 0.28
HNM	18.0	9.50	0.30	3.50	0.50	-	-	-	-	P 0.25
A286	15.0	25.0	0.06	1.20	0.50	-	1.20	2.00	0.25	V 0.30

Metallurgically stabilizing the composition in such stainless steels is a very critical issue. A small fluctuation that may occur in the composition may cause delta ferrite formation to a large extent in the solution taking-up attenuation. If the austenite is very stable, too much austenite may remain in the interior at room temperature after the precipitation. These two events prevent all hardening during aging, which

significantly affects carbon and nitrogen content. For example, species such as 17-4 PH and stainless W contain some ferrite arrays in the martensitic structure, while species such as 15-5 PH and custom 450 do not contain ferrite after rapid cooling.

The appearance of precipitates or the appearance of a tempering effect depends entirely on the steel composition. Semi-austenitic precipitation hardened stainless steels are under the influence of the M_s temperature, solution annealing and compositions. For example; AM 350 steel, when solving solution is applied at he temperature is below 930°C , the temperature of the resultant M_s increases above the room temperature when the carbides do not completely pass through the solution. When it is above 930°C , the M_s temperature also decreases suddenly. In practice, delta ferrite formation occurs at high temperatures and it is recommended not to exceed 1050°C . Austenitic precipitation hardened stainless steels contain alloying elements to stabilize the austenitic internal structure even after annealing and any aging or hardening, even after cold forming [17].

The heat treatments for the austenitic precipitation hardened stainless steels are summarized in Table 2.5. The phase forming the settling cure passes through the solution at temperatures of $1100\text{-}1300^{\circ}\text{C}$ and remains in solution in the austenitic structure even after rapid cooling from these temperatures. In a long-term aging process in the temperature range of $650\text{-}770^{\circ}\text{C}$ following the dissolution process, the elements such as titanium, aluminum and phosphorus form metallic compounds to increase the hardness and strength of the austenitic structure. The structure retains its non-magnetic property, although the hardness obtained is lower than the values obtained in martensitic or semi-austenitic precipitation hardened stainless steels. All austenitic precipitation hardened stainless steels generally retain their austenitic internal structure during shaping, welding and heat treatment. However, some alloying elements added to effect precipitation hardening considerably affect their welding capabilities [17].

Table 2.5. The heat treatments for the austenitic precipitation hardened stainless steels.

Species	Receive Resolution		Aging	
	Temperature (°C)	Cooling Medium	Temperature (°C)	Time (Hour)
Mantenzitic				
17-4 PH	1050	O. A	480 or 500-630	1
15-5 PH	1050	W	480 or 500-630	1
Custom 450	1050	W	480-630	4
Custom 455	840	W	480-570	4
Stainless W	1050	A	510-570	0.5
Semi Austenitic				
17-7 PH	960	A	(1)68	8
PH 15-7Mo	760	A	(2)510-600 or 570-600	1 or 1,5
PH 15-7 Mo	950	A	(1)68	8
PH 15-7 Mo	950	A	(2)510-570	1
AM350. or AM355.	940	A	(1)73	3
AM350. or AM355.			(2)460-540	3
Austenitic				
A-286	990	O. A	725	16
17-10P	1130	W	710	24
HNM	1130	O. A	740	16
O-Oil, A-Air, W-Water				

2.3.4. Duplex Stainless Steels

The most important advantages of such stainless steels having a double-phase internal structure and are much better corrosion resistance. The use of chlorine-containing fluids in industrial applications requires that the materials to be used have increased strength and improved corrosion resistance compared to known stainless steel grades (316L X2CrNiMo17133). Such steels become superplastic when their grain size can be reduced to 3-10 μm , showing a tensile strain of 500% at 950°C [19].

Compared with ferritic chromium stainless steels containing 18% or more chromium, the main austenitic chromium-nickel stainless steel grades exhibit fairly limited stress corrosion crack resistance in chlorine-containing environments. However, austenitic chrome-nickel stainless steels have better weldability. Ferritic chromium stainless steels are resistant to stress corrosion cracking, but they are not well

weldable, including the recently developed grades. This is because during the welding, there is an increase in grain in the heat affected zone. As is known, many types of ferritic chromium stainless steels contain 0.1% and more carbon, which promotes the formation of intergranular brittle martensite structure. Both of these events lead to a drop in ductility and cold cracking of the weld, so duplex stainless steels have been developed to combine the good relationship between microstructure and mechanical properties [19].

The first duplex stainless steel, an alloying failure resulting from the production of 18% chromium, 9% nickel, 2.5% molybdenum stainless steel in the J.Holtzer Company in France in 1933, is found by chance, resulted in the formation of a stainless steel with high ferrite content in the austenitic matrix. In the following years researches on this subject were continued and various patents were obtained. Between 1930 and 1940, studies on the development and industrial production of duplex stainless steel were carried out in Sweden and the U.S. [14-17].

In fact, it is understood from the scientific studies on duplex stainless steel that the first commercial duplex stainless steel was produced in the 1970's and introduced to the market. The chemical composition of this steel was approximately 18.5% chromium, 5% nickel, 2.7% molybdenum and 0.1% nitrogen. However, this quality has not attracted much attention to users in the future. Because of its high carbon content, weldability is very poor. Many duplex stainless steels with very good weldability and high corrosion resistance have been developed in the last decade by reducing the amount of carbon produced in the steelmaking technology, and many researchers are now working on the weldability of these species, the enhancement of the mechanical properties of welds and the further enhancement of corrosion resistance [17].

Due to its superior features, duplex stainless steels are offered to the industry in different shapes and sizes and are widely used in heat exchangers, oil, gas and sea water pipes and fittings, marine oil platforms, gas wells, carrier cups, casting pumps and valve bodies, ship propeller and in parts construction, geothermal applications. For example, X2CrNiMoN2253 (DIN 1.4462, UNS-S31803, or SNS205 according to

the new UNS) steel has an increasingly widespread application area in the chemical industry, due to the fact that it is resistant to 317L for steel, pitting and gap corrosion, and 304L and 316L for stressed corrosion [18].

Duplex stainless steels have two distinct phases in a simple sense: ferrite and austenite. They are therefore known as ferritic-austenitic or austenitic-ferritic stainless steels. As already mentioned, they have the best common properties of austenitic stainless steels and ferritic steels. With ferritic structure, resistance to tensile and stress corrosion cracking, with austenitic structure, toughness and general corrosion resistance are ensured. Thus, a two-phase fine-grained, high strength and good corrosion resistant steel is produced. Modern duplex stainless steels suffer from loss of toughness below -50°C in the heat treated condition. As a result of the presence of the ferrite phase in these steels, the ductile-brittle transition temperature decreases with decreasing temperature in the notch impact curve. Together with the 475°C brittleness, this factor has a limiting effect on the use temperatures of duplex stainless steels. Therefore, such steels maintain their mechanical properties very well between temperatures of -50 to 280°C . When resource availability and comparable prices are added to this, today's indispensable types of stainless steels are emerging [17].

The main alloying elements in the contents of duplex stainless steels are chrome, nickel, molybdenum and nitrogen. While chromium and molybdenum balance ferrites, nickel and nitrogen balance austenite. Additionally, some species include manganese, copper or wolfram. As a result of research on duplex stainless steels, new steels with 22-27% Cr, 4-8% Ni, 0.1-0.3% Ni have been developed, which contain 1.5-4% molybdenum or no molybdenum and which contain additional alloying elements in some conditions. These are used as super and duplex stainless steels [18].

Duplex stainless steels have an internal structure containing approximately 50% ferrite and 50% austenite. However, in order to achieve this structure, it should be suited to the chemical composition and heat treatment very well. In rolling or cast duplex stainless steels, the microstructure is usually obtained by a heat treatment

applied at a temperature range of 1040-1150°C. In casting, such stainless steels usually contain 80% or more ferrite and a small amount of austenite is formed. In addition, brittle metal phases such as σ and / or χ phases are common in the world. If the heat treatment temperature is high, it prevents the formation of metallic compounds and the microstructure at room temperature usually contains 40-60% ferrite with austenite stabilized. In addition to the tendency to form metallic compounds during slow cooling or at a temperature range of 540-930°C, duplex stainless steels are also exposed to a metallurgical phenomenon called at 475°C (temper) embrittlement. This brittleness is occurred due to the precipitation of iron rich ferrite (α), chromium rich ferrite (α'). Duplex stainless steels are not recognized much in our country and there is no Turkish Standard for such stainless steels. Despite the preparation of standards for such steels in western countries, it is often recognized by the brands of the manufacturer companies. The identification codes and chemical compositions of these steels used in various countries are also given in Table 2.6. The mechanical properties of such stainless steels depend on the production process (rolled, cast) and final heat treatment, as well as on all steels [17].

Table 2.6. The identification codes and chemical compositions of duplex stainless steels used in various countries.

Common Name	UNS No	BS EN No	Steel Type	Typical Chemical Composition %						0.2% proof strength N/mm ² (min)
				%C	Cr	Ni	Mo	N	Cu	
2304	S32304	1.4362	duplex	0.015	23.0	4.0	0.055	0.13		400
2205	S31803	1.4462	duplex	0.015	22.0	5.5	3.0	0.14	-	450
2205	S32205	1.4462	duplex	0.015	22.5	5.5	3.3	0.17		450
255(UR52N)	S32520	1.4507	super duplex	0.015	25.0	7.0	3-5	0.28	0.13	550
2507	S32750	1.4410	super duplex	0.015	25.0	7.0	4.5	0.28	0.3	550
Zeron 100	S32760	1.4501	super duplex	0.015	25.0	7.0	3.5	0.25	0.8	550
Sandvik SAF3207	S33207	-	hyper duplex	0.03	31	7.5	4.0	0.50	0.75	700

2.3.5. Austenitic Chrome-Nickel Stainless Steels

In the iron-chromium binary equilibrium diagram (Figure 2.3), it is seen that except for the region containing 13% excess chromium, the austenite area is not observed and at every temperature range, the structure is ferritic and only a narrow $\alpha + \gamma$ region is found in the chromium range of 12-13%. This ferrite is called δ - ferrite because it normally occurs when the steel is solidified from the liquid state. It has been previously shown that the addition of carbon to these two alloying causes the γ area and especially the $\alpha + \gamma$ area to expand. Adding carbon up to 0.4% to a steel with 18% chromium does not preclude the internal structure from remaining completely ferritic and not allowing the conversion to take place, however in the range of carbon content of 0.08-0.22%, the structure shows partial conversion and the internal structure of $\alpha + \gamma$ can be obtained. If it contains more than 0.4% carbon, steel can be cooled rapidly from the zone γ and kept in a completely austenitic structure at room temperature. Another effect of carbon in the internal structure also occurs in the formation of carbides (Figure 2.9) [20].

In austenitic stainless steels, $M_{23}C_6$ carbide is the most important carbide formed, which has a significant effect on the corrosion resistance of the steel. Nickel incorporated in the alloy containing low carbon and 18% chromium expands the γ phase formation zone, and when the nickel amount reaches 8%, the γ area goes down to room temperature. This phenomenon led to the birth of a steel group containing 18% chromium and 8% nickel, the most well known austenitic steel species.

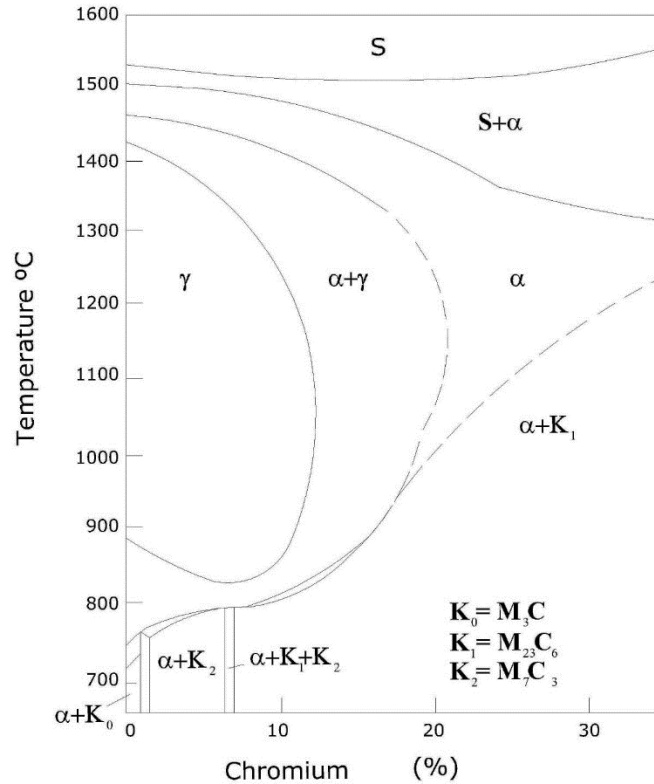


Figure 2.9. Effect of carbon on iron-chromium balance diagram (0.05% C) [20].

This special composition is able to keep the inside of FCC in a balanced state with minimum nickel content at room temperature. Because the chromium content slightly reduction or multiplication requires more nickel to remain in a stable austenite phase. For example, stainless steel with 25% chromium, which is more resistant to corrosion, requires 15% nickel to keep it in the austenitic form at room temperature.

As is known, obtaining iron alloys in an austenitic structure has been one of the most important efforts of metallurgists since the beginning of this century. Austenitic chrome-nickel stainless steels, now called the AISI 300 series, are the products of these studies. Today austenitic stainless steels contain elements such as molybdenum, tantalum, titanium, niobium, and nitrogen which have been incorporated to improve 16 to 26% chromium, 10 to 24% nickel, up to 0.4% carbon and some other properties. The amount of austenite-forming elements has been further increased in groups that are developed in recent years, called full or super austenitic stainless steels [21].

Austenitic stainless steels can not be hardened by quenching as there is no austenite ferrite transformation during cooling. Such non-magnetic stainless steels are symbolized as being in high alloy steels according to TS 2535, EU 95, EU 88, and DIN 17440 and as well as being grouped in the AISI 3XX series. In TS 2535 austenitic stainless steels are defined as 'stainless steels' are intended to provide a chrome and austenitic structure against corrosion with nickel in the composition, which are not magnetic in room temperature, can not be cured by heat treatment suitable for cold forming. Typical austenitic chrome-nickel stainless steels are given in the shortened views according to AISI and DIN norms [18].

The corrosion resistance of austenitic chrome-nickel stainless steels is higher than that of martensitic chromium and ferritic chrome stainless steels. The austenitic nature of the inner structure and embrittlement under transition temperature which is a very important problem encountered in ferritic stainless steels can't be seen in these types of steels. Both the subzero (up to -270°C) or high temperature corrosion resistance, and the superior mechanical properties of this steel group make it possible to use it as an unrivaled structural steel in many areas [18].

X5CrNi1810 (304) stainless steel is an austenitic stainless steel widely used for corrosion resistance and the ability of good formability. This type of steel is also used frequently in the construction of machine parts which require high strength since the strain hardening is also increased. X10CrNi188 (302) steel is high carbon and has high strength with cold strain hardening. The X10CrNiS189 (303) steel has been converted into veneer steel with sulfur additive and is used for making stainless bolts, spindles and valves. X2CrNi1911 (304L) steel is a type of carbon reduced X5CrNi1810 (304) steel. The aim of the development of these steels, known as very low carbon (ELC) stainless steels, is preventing carbide precipitation, which occurs in the heat affected zone during welding and causes intergranular corrosion [21].

X15CrNiSi2012, X5CrNiSi2520, X15CrNiSi2520 steels are corrosion and crack resistant steels with high nickel and chromium content at high temperatures and are used in combustion chambers and high temperature working parts. The steels X5CrNiMo1722 (316), X2CrNiMo17132 (316L) containing molybdenum are more

widely used than the steel X5CrNi1810 (304) in the marine and chemical industries. The X2CrNiMo17132 (316L) steel is also manufactured with low carbon content to prevent grain corrosion problems. The steels X6CrNiTi1810 (321) and X6CrNiNb1810 (347) were stabilized with titanium and niobium to prevent the tendency for intergranular corrosion at welds at high temperatures. Austenitic chrome-nickel stainless steels are often used in damp environments. Increasing molybdenum and chromium contents play a role in increasing corrosion resistance against corrosive solutions. The high nickel content reduces the risk of stress corrosion cracking. The resistance of austenitic stainless steels to general corrosion, cavitation and crevice corrosion increases depending on the amount of alloying elements (in particular chromium and molybdenum) involved [17].

Table 2.7 shows the tensile and notch-to-impact strengths of a group of austenitic stainless steel at room temperature. The effect of chemical composition on these properties during the deformation can be monitored immediately. For example, since the yield strength is generally the end point of the elastic deformation, there is no effect of plastic deformation up to this point. The effect of the composition on the flow limit is simply the effect of the solid solution curing, and the elements that are most effective in this solid solution curing are the interstitial elements such as carbon and nitrogen. In this respect, high yield strengths of steels containing carbon and nitrogen are higher.

Table 2.7. Mechanical properties of some austenitic stainless steels at room temperature [17].

Steel*	Product Format	Flow** Strength MPa	Tensile Strenght MPa	Elongation (50 mm'de %)	Shrinkage %	Notch-Impact (min)J	Modulus of Elasticity MPa
S20100 (201)	Sheet or Band	310	655	40.0	-	-	197×10^3
S20200 (202)	Sheet or Band	310	621	40.0	-	-	-
S30100 (301)	Sheet or Band	276	758	60.0	-	-	197×10^3
S30400 (304)	Sheet or Band	290	580	55.0	-	-	197×10^3
S30400 (304)	Plate and Rod	241	565-586	60.0	70.0	149	197×10^3
S31000 (310)	Plate	310	655	50.0	65.0	122	197×10^3
S38100*** (XM15)	Plate	207	517	40.0	-	135	200×10^3
* (UNS number (AISI number)). ** %0.2 limit *** USS 18-8-2							

It has previously been shown that high degree of deformation hardening allows these steels to have very high yield and tensile strength, and it is interesting here that they protect some of the ductility and toughness of such steels even in these high yield and tensile strengths. In this respect austenitic stainless steels are high strength and high corrosion resistant steels cold rolled or drawn. In the case of natural welding, all or part of the part will lose properties acquired by cold deformation. But with a well-thought-out welding method, such stainless steels are easily combined. Even in practice, all the effects of cold deformation on strength are utilized. Examples of best practices in this regard are railway vehicles, trucks, trailer casings where cold-deformed austenitic stainless steels are used. Figures 2.10 and 2.11 exhibit cold-drawn tensile strength properties of the AISI 202 and 310 steels respectively. When the shapes are examined, the metastable steels 202 for a certain rate of deformation; have higher yield, tensile and elongation % than stable 310 [21].

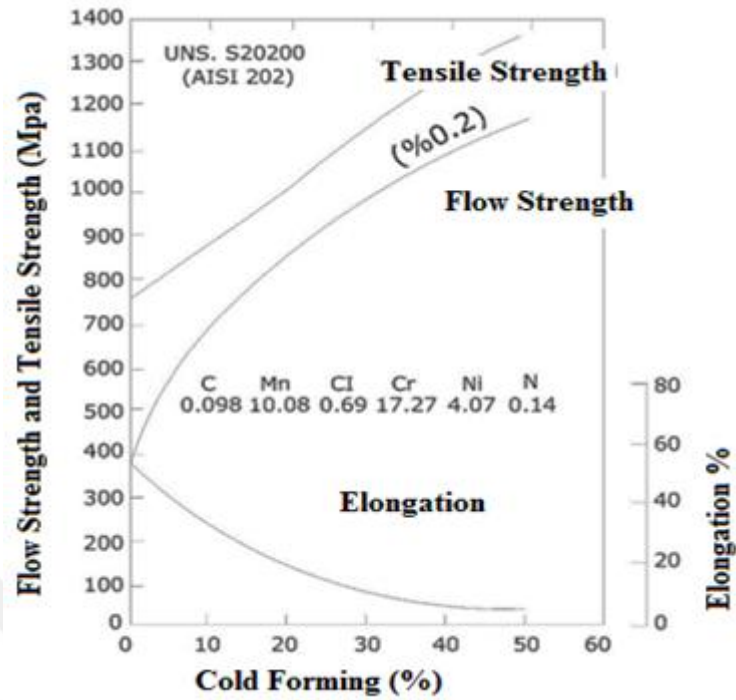


Figure 2.10. Effect of cold deformation on mechanical properties of AISI 202 stainless steel [21].

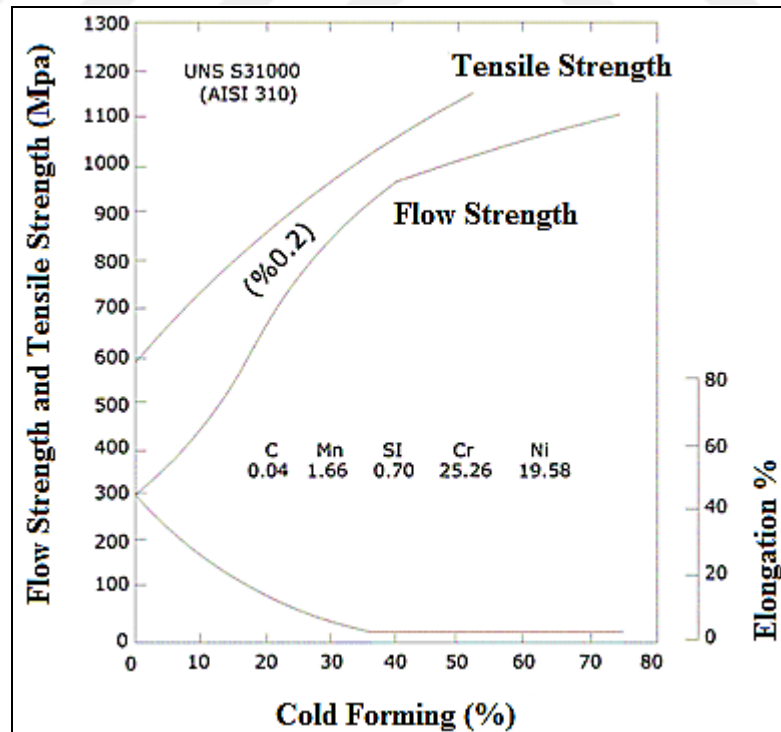


Figure 2.11. Effect of cold deformation on mechanical properties of AISI 310 stainless steel [21]

The chemical composition of austenitic stainless steel influences its properties. The composition first effects mechanical properties, i.e., the stability of the steel, and secondly the solid solution strengthening. The inclusion of almost all of the alloying elements makes austenitic more stable and the effect of this composition is more complex. For example, the incorporation of an element may initially have little effect on stabilizing the austenitic structure, whereas a high solid melt hardening effect results in high yield and tensile strength and lower elongation. On the other hand, the greater the participation of this element, the more intense the austenite balancing effect. As a result, yield and tensile strength are reduced, but higher elongation is obtained. In addition, the temperature at which the process is performed is of great importance. As cold rolling occurs below the M_s temperature, the plastic deformation will cause austenite to transform into martensite and increase in strength at significant levels. When cold deformation occurs at around M_s , martensite will not form and deformation hardening will occur at a lower rate. In metastable austenitic stainless steels the M_s is at room temperature and therefore in these species there is a very high strength increase in the cold rolling finish at room temperature.

For example, if the rolling is carried out at 200°C in these species, the deformation hardening is observed to fall immediately and consequently the energy consumed in the process at these temperatures is much lower. In those cases where the cold rolling power is limited in the case of steels, the temperature is slightly increased and a considerable advantage is achieved. In super-austenitic stainless steels the austenitic structure is continuous and stable and under all conditions is completely austenitic. After welding, no ferrite is formed in the weld metal as opposed to AISI 3XX. Full-austenitic chrome-nickel stainless steels are alloyed at a very high rate in order to increase their corrosion resistance, and in some respects, they are called super-austenitic stainless steels in some literature (Table 2.8). These super-austenitic stainless steels have been developed for use in the chemical industry and to prevent corrosion problems in materials operating in extreme corrosive environments [21].

Table 2.8. Super austenitic stainless steels and chemical compositions [21].

Chemical Composition	C max	Cr	Mo	Ni	Other
18Cr/16Ni/3,5Mo/L	0,030	17,0-19,0	3,0-4,0	14,0-17,0	-
17Cr/13Ni/4,5Mo/L	0,030	16,5-18,5	4,0-5,0	12,5-14,5	N=0,12-0,22
24Cr/20Ni/Nb/L	0,030	23,0-25,0	-	19,0-21,0	Nb=> 10X%C
25Cr/22Ni/2Mo/NL	0,030	24,0-26,0	2,0-2,5	21,0-24,0	N=0,10-0,16
23Cr/27Ni/3Mo/3Cu/NbL	0,030	22,0-24,0	2,5-3,0	26,0-28,0	Cu=2,5-3,5, Nb=0,6-0,8
20Cr/25Ni/6Mo/2Cu/NL	0,030	19,0-21,0	5,5-7,0	24,0-26,0	Cu=1,0-2,0, N=0,13-0,17

Special super austenitic stainless types have been produced for such applications as sulfuric acid, phosphoric acid, acetic acid, chemical fertilizer production. In particular, the corrosion resistance of these steels (PRE: Pitting Resistance Equivalent) and the improved critical pitting temperature (CPT) have been increased. These types are widely used in seawater environments for a variety of purposes, and even at high temperatures range corrosion resistance is very good. In addition, super-austenitic stainless steels are not magnetic and have a superior toughness at subzero temperatures. The super-austenitic stable structure is achieved by means of high nickel content or nitrogen content [21].

2.3.5.1. Type 316L Austenitic Stainless Steel

316L austenitic stainless steel is an extra low carbon version of the 316 steel alloy. The low carbon content at 316L is the least damaging to harmful carbide precipitation in the welding. As a result, 316L is used when welding is required to ensure maximum corrosion resistance. Although type 304 is similar, both types 316 and 316L have better corrosion resistance and are stronger at higher temperatures. They can not be hardened by heat treatment and can be easily drawn and form. Annealing 316 and 316L stainless steels require heating between 1038-1149°C before quick extinguishing [22]. Physical properties and uses of 316L austenitic stainless steel as follows;

Physical Properties:

- Density: 0.799g/cm³
- Electrical resistivity: 74 microhm-cm (20°C)
- Specific Heat: 0.50 kJ/kg-K (0-100°C)
- Thermal conductivity: 16.2 W/m-k (100°C)
- Modulus of Elasticity (MPa): 193×10^3 in tension
- Melting Range: 1371-1399°C

Uses of 316Lstainless steel alloy:

- Architecture,
- Bridges,
- Automotive bodies,
- Aircraft,
- Airports,
- Surgery and dentistry
- Kitchens,
- Power plants [22].

CHAPTER 3

STRAIN AGING

3.1. INTRODUCTION

It is generally accepted that the strain ageing is due to the diffusion of nitrogen and / or carbon atoms in the solution to dislocations created by plastic deformation. Initially, an atmosphere of carbon and nitrogen atoms is created that is immobilized it along a dislocation length. However, extended ageing results in sufficient carbon and nitrogen atoms for the precipitates to form over the dislocation length. These precipitates interfere with the movement of subsequent dislocations, resulting in loss of ductility and some hardening. A thermally activated process, the stress-strain rate, is mainly dependent on the temperature and ageing time. Generally, prolonged ageing results in a satiety value with no effect of further aging. In fact, ageing at very high temperatures can lead to excessive aging, accompanied by a gain in ductility and a loss of hardness. A greater increase in strain ageing is usually provided by repeated aging and strainining. Free carbon and nitrogen atoms should be present in the solid solution for tension to appear, but very little carbon and nitrogen are required. For example, it is sufficient to produce as little as 0.01 percent [23].

The two types of strain ageing, dynamic strain ageing and static strain ageing that can occur during or after deformation. Common properties that harden the material can also increase the ductile-brittle transition temperature and ultimate tensile strength and reduce ductility. Carbon and nitrogen are generally responsible for strain ageing because they have a relatively fast diffusion in steels, even at very low temperatures, between 200°C and 400°C.

Because of the spreading forces due to a strong temperature function, the level of strengthening depends on the ageing temperature and time. It becomes more complicated when the temperature rises above 800°C due to events such as recrystallization and recycling phenomena. As recrystallization and recovery removes displacement all together, it will decrease the occurrence of strain ageing and change the mechanical behavior [24].

3.2. STATIC STRAIN AGEING

The behavior of steel is significantly affected by the method of manufacture and its chemical composition. Bauschinger (1887) observed an increase in the maximum load carrying capacity of a steel piece after it was tested in the inelastic range, after being left for a while and retested. This phenomenon, known as strain-ageing, can be interpreted by the time-dependent propagation of small elements such as a nitrogen and carbon on crystal defects (such as dislocations). The presence of small atoms (such as nitrogen and carbon) can significantly increase the shear resistance between atoms, and thus increase yield strength, while plastic deformation is triggered by the movement of dislocations through the crystals. Over the past 50 years, a wide range of research has shown that all structural steels are potentially sensitive to strain-ageing. Strain ageing causes an increase in ultimate tensile strength and yield strength of reinforcing steel bars as well as a lower tensile elongation capacity is observed in strain aged steel bars [23].

Static strain ageing occurs after the plastic deformation of a material has passed and the material is aged for a period of time. The reason for static strain ageing is that soluble atoms are placed on dislocations and dislocations are pinned effectively. Nitrogen and carbon, which are capable of generating ageing effects in relatively short hold times and at low temperatures, are generally responsible for static strain ageing. This is because these interstitial elements have significantly higher diffusivities than the substitution elements. Nitrogen has a higher diffusion coefficient and solubility in steels, which gives a more pronounced effect on strain ageing. Static strain ageing can lead to flow localization during cold working [24].

When steel has been strained (deformed plastically) and then allowed to age, it has been subjected to what is known as static strain ageing. Figure 3.1 shows the stress-strain curve for mild steel (low carbon steel) compared to those pre-strained to the same steel (point B) and then exposed to ageing. The tensile strength and yield strength in the static strain ageing result increased ΔY and ΔU , respectively, and the elongation amount, Δe , was decreased. In general, static strain ageing results in the loss of some ductility but increases strength very little. Static strain ageing is known to be accompanied by a decrease in the fracture toughness and which is caused an increase in the brittle transition temperature (BTT) [23].

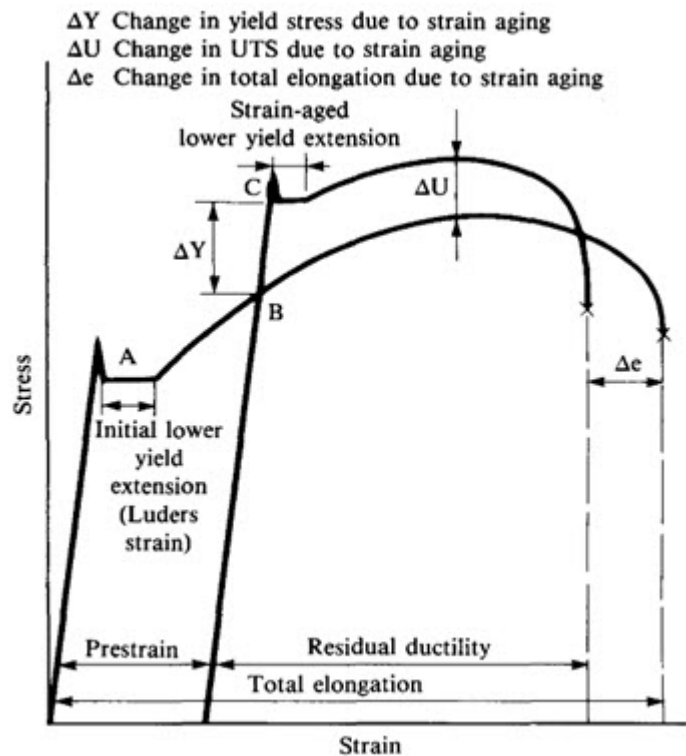


Figure 3.1. A schematic representation of the influence of static strain ageing on the stress-strain curve for mild steel [23].

3.3. DYNAMIC STRAIN AGEING

3.3.1 General Properties

It is known that atoms dissolved in metals and alloys at sufficiently high temperatures are fixed by displacement interactions or in other words by the

substitutional atoms and thus the stress required for the displacement movement is increased. Generally, as a result of ageing, there is an increase in the material hardness and the yield strength. The ageing process occurs slowly at room temperature and takes place faster at higher temperatures because it helps to diffuse the ageing elements as temperature increases. This phenomenon is often referred to as dynamic strain ageing which occurred in ageing processes in alloys containing soluble atoms that can be quickly and strongly separated into dislocations and locked during the strain [25].

The maximum effect of dynamic strain ageing corresponds to these conditions in which diffusions can monitor changes in the displacement structure of dissolved atoms. Known physical specifications of dynamic strain ageing are serrated plastic flow, blue brittleness, negative strain rate sensitivity, strain hardening with temperature and maximum in the variation of ultimate tensile strength and Hall-Petch reaches its peak when the curve changes with temperature. Increased strain hardening rate and deformation localization observed during flow of nonhomogeneous plastic affect the mechanical performance of materials in terms of break-up time and ductility in low-cyclic fatigue mode [26].

Dynamic strain ageing is an effect of the interaction between dissolved atoms and moving dislocations that occur during deformation. The appearance of the dynamic strain ageing strongly depends on the rate of strain and temperature, if the stress rate drops or the temperature rise dynamic strain ageing is condensed. Dynamic strain ageing arises when dissolved atoms provide sufficient mobility to keep pace with mobile dislocations and create dissolved atmospheres. Dynamic strain ageing occurs during viscous slip-type dislocation movement or during temporary dwells in local obstacles which contributes to matrix hardening and results in increased resistance to dislocation motion.

For this reason, an increase in flow stress is required to maintain the applied tensile ratio in order to create new dislocations or unlock the dislocations. Both of these causes an increase of the mobile dislocation density. A typical character of dynamic strain ageing, often called the Portevin-LeChatelier (PLC) effect, is the formation of

serrated yielding. Dynamic strain ageing may also cause an increase in flow stress and work hardening rate, and a most significant negative stress rate sensitivity [4]. The instantaneous increase in plastic strain caused imbalances in plastic strain or in other words serrated flow. Figure 3.2 shows the load-extension graphs of Nb microalloyed steel, which is subjected to tensile tests at 25°C to 450°C after austenitisation at 1000°C for 1 hour and then air cooled. Serrated flow, one of the characteristics of dynamic strain ageing, was clearly observed at certain temperatures in all specimens. The serrations on the stress and strain diagrams started to appear in the workhardening zone at 200°C, and as the temperature increased to 250°C, the serrations increased in size and frequency. At higher temperatures, the serrations began to slip from the curves until they were mostly lost at 350°C [27].

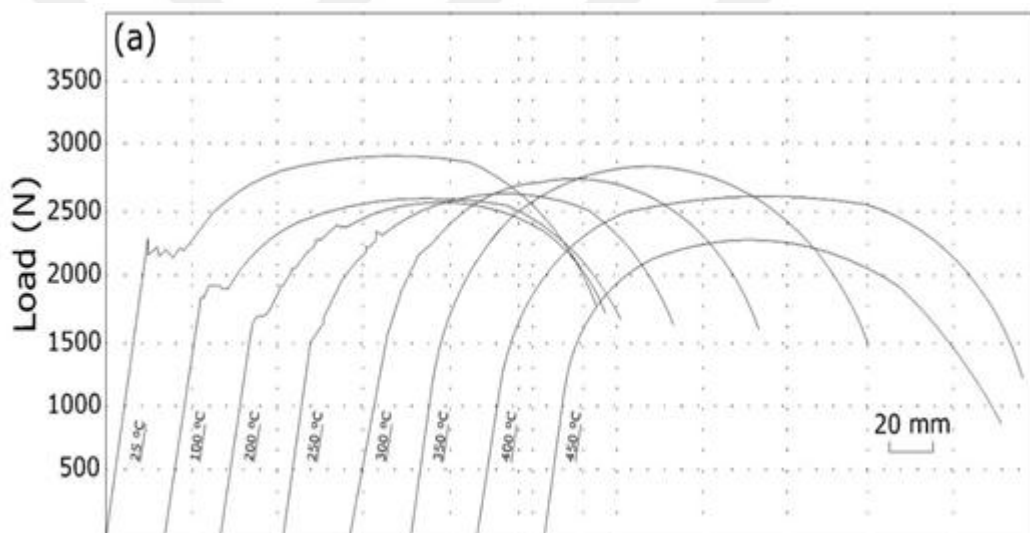


Figure 3.2. Load–extension curves for Nb microalloyed steel tensile tested at given temperatures [27].

3.3.2. Classification and Morphology

In the dynamic strain ageing temperature range, the plastic flow becomes unstable. This is manifested by the flow of intermittent (or serrated) plastic in the stress-strain curve. Portevin and Le-Chatelier were first investigated these irregularities in detail in aluminum alloys [20, 28]. It is now called the Portevin-Le Chatelier effect (PLC) to be associated with these cracks. Generally, there are two types of instability related to strain ageing: Portevin-Le Chatelier (PLC) and the Lüders front. The

Lüders column is the line between the plastic deformation and the unmodified material in the drawing case [29]. If the crosshead speed of the test machine is held constant toward the other end, it is located at one end of the specimen and is typically spreaded constant speed [26]. The nominal stress-strain curve appears smooth during diffusion. However, the localization comes before the yield point behavior; After reaching a summit, the flow stress drops quickly to a lower value. On the other hand, the PLC effect appears as a series of shear bands that appear regularly at regular intervals, or as a series of spreading bands with a source at one end of the sample [30]. The resulting nominal stress-strain curve is serrated; plastic strain and stress oscillate. During oscillations, the average stress can be increased or increased constantly, or by specifying continuous tensile stiffness at the steps corresponding to the intersection of the sample lengths by the bands. These serrations may be of several types. Five different plastic discontinuities were reported for dynamic strain ageing as A, B, C, D and E as seen in Figure 3.3.

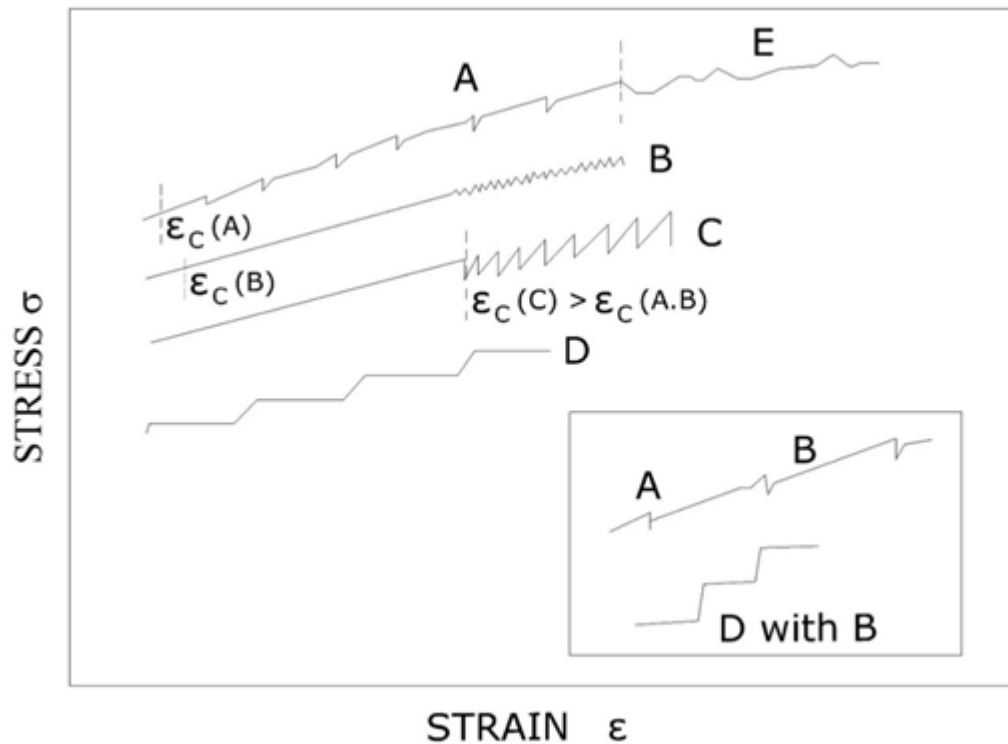


Figure 3.3. Stress-strain curves illustrating the various types of serrations [26].

A-type serrations are periodic indentations caused by the repeated nucleation of a cutting band at the peak tension and subsequent spreading of the band continuously

along the measuring length in the direction of tension. These are generally thought of as locking fingers characterized by a sudden rise following a falling stress below the level of the stress-strain curve. These occur at the low temperature (high strain rate) of the dynamic strain ageing regime.

B-type serrations, which are oscillations in the general level of the rapid successive stress-strain curve, result from the nucleation of narrow cutting bands at intermittently scattering or non-spreading peak stresses, and are then caused by nucleation of another cut, either in an adjacent section or at some distance from the first B-type serrations can develop at high tensile strengths from A-type or can start with plastic flow immediately at higher temperatures and at lower strain rates than in the A-type case. In practice, it can be mixed in combination with notches A and B type [26].

C-type serrations are stress drops that occur below the general level of the flow curve, and are thought to be caused by the opening of the dislocation. At higher temperatures and at lower strain rates it is more than A and B type serrations.

D-type serrations are platelets in the stress-strain curve, also called 'ladder type' that occurs due to shear band spreading without work hardening or strain gradient in front of the motion band. D type serrations can also occur in B type mixed mode. E-type serrations are similar to A-type, but do not cause hardening of the work during spreading of the tape. The plastic flow imbalances of the E-type are seen in A-type high tensile strengths. When the test is carried out in the stress control mode, the type D serrations can be observed (the load is kept constant or increased by a constant rate) [18].

3.4. THE ROLE OF CARBON AND NITROGEN ON STRAIN AGEING

It is often believed that the strain ageing process is caused by the diffusion of carbon or the nitrogen atoms in solution due to the dislocation of plastic deformation. Rightly where the atmosphere of the carbon and nitrogen atoms were composed along with the dislocation length that makes it fixed. However, the

prolonged ageing can lead to enough carbon and nitrogen atoms to form an accretion along the length of the dislocation.

These deposits prevent subsequent dislocation movement and cause some hardening and ductile loss. The degree of strain ageing is a thermal activation process, which mainly relies on the ageing time and temperature. In general, extended ageing leads to full values, and further ageing does not affect it. In fact, ageing at very high temperatures can lead to excessive ageing, loss of hardness and increased ductility. The repeated strain and ageing always lead to an increase in the ageing of the strain ageing property itself, free nitrogen and carbon atoms of the solid solution ought to be available, but a less degree and amount of carbon and nitrogen are required. For example, only 0.02 percent is enough to produce [31].

Alloying elements in steel are diffused in their own microstructure components, mainly iron and iron carbide. In the situation of nitrogen and some carbon that were, however, not absorbed by the iron carbide, were in the iron-rich phase looking like; small individual atoms in the interstices of the crystal form. Over time, as the steel calms down, the carbon atoms and nitrogen atoms drift to dislocations through the structure due to the deformations that occur in the lattice.

These (small) interstitial atoms give a stabilizing effect on dislocation motion that raises the force required to cause dislocation slip. It now needs more power to distort the steel and increase its strength. If both nitrogen and carbon are present, iron-carbon-nitrogen compounds (carbonitrides) can be formed which increase the strength of the steel and also limit the movement of dislocations.

The effect of temperature is significantly vital and relevant for the "ageing" of structural steels. Structural steels are more complicated because they contain relatively more alloys and carbon and have broader microstructures. Because of that, the ageing of the steel is quantified by the raise in strength and loss of the ductility that does not happen at the room temperature. Mainly, temperatures in the 200°C to 400°C range of 2 to 6 hours are required for ageing.

When cold deformation (alone) is applied to the steel, a second strengthening mechanism appears. When the dislocated interstitial atoms disengage to cause slippage, they begin to cross each other. The complex series of interactions that happens between the dislocations immobilizes them and reduces their mobility. The lower liquidity also results in the higher strength, the lower toughness and the lower ductility. Due to these occurrences, the cold-deformed steel has some reduced ductility and toughness prior to any strain ageing that happened and the ductility and the toughness losses were greater with the cold-deformation heating. It is this type of structural toughness that damages the most serious incidents. These were the two effects that raises the strength and ductility and the ductility due to cold strain, as well as increased strength and loss of ductility as ageing is a major factor in strain ageing [32].

3.5. SOLUBILITY OF CARBON AND NITROGEN IN FERRIT AND AUSTENITE

The atomic radii of carbon ($0.8\text{\AA}$) and nitrogen ($0.7\text{\AA}$) are much smaller than that of iron ($1.28\text{\AA}$), which allows these solute elements to enter into the interstices or “holes” of the α -iron and γ -iron crystal lattices. On a hard sphere model for fcc austenite, the largest hole at an octahedral site is $0.52\text{\AA}$ in radius which is surrounded by six atoms located at the corners of a regular octahedron (Figure 3.4a); the next largest hole is $0.28\text{\AA}$ in radius at a tetrahedral site which is surrounded by a tetrahedron of four atoms (Figure 3.4b). In spite of the fact that the bcc ferrite is not densely packed, the octahedral site is only $0.19\text{\AA}$ in radius and is surrounded by six atoms at the corners of a slightly compressed octahedron (Figure 3.5.a) while the tetrahedral site is $0.36\text{\AA}$ in radius (Figure 3.5b). Evidently, octahedral holes in α -iron are much smaller than those in γ -iron. Since carbon and nitrogen are larger than the available interstices in either lattice, it is apparent that some distortion must ensue when they occupy the interstices of iron lattices. However, these interstitial atoms reside in octahedral sites, with an expansion caused by the displacement of two nearest- neighbor iron atoms rather than four, as in tetrahedral sites.

Table 3.1 shows the solubility values of carbon and nitrogen in γ -iron and α -iron. As can be seen that the solubility of nitrogen (2.8 wt.%) and carbon (2.04 wt.%) in γ -iron is much greater than α -iron as a result of ostenite much larger octahedral holes. This large difference in solubilities is fully exploited to improve strength. Nitrogen which has great significance in the heat treatment of steels has more solubility than carbon in ferrite. That is why the nitriding heat treatment is usually performed at a temperature lower than the eutectoid temperature [33].

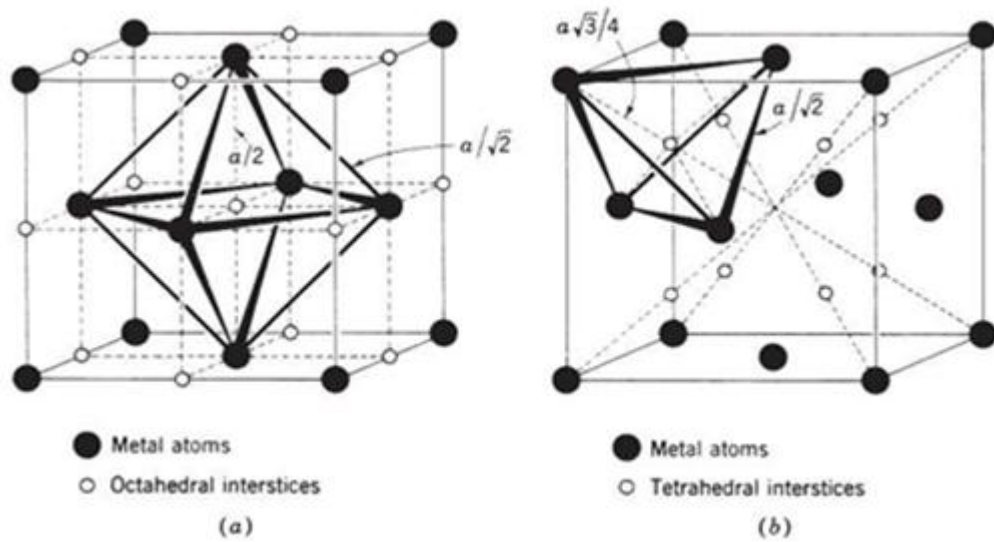


Figure 3.4. (a) Octahedral and (b) tetrahedral interstitial voids in fcc structure [34].

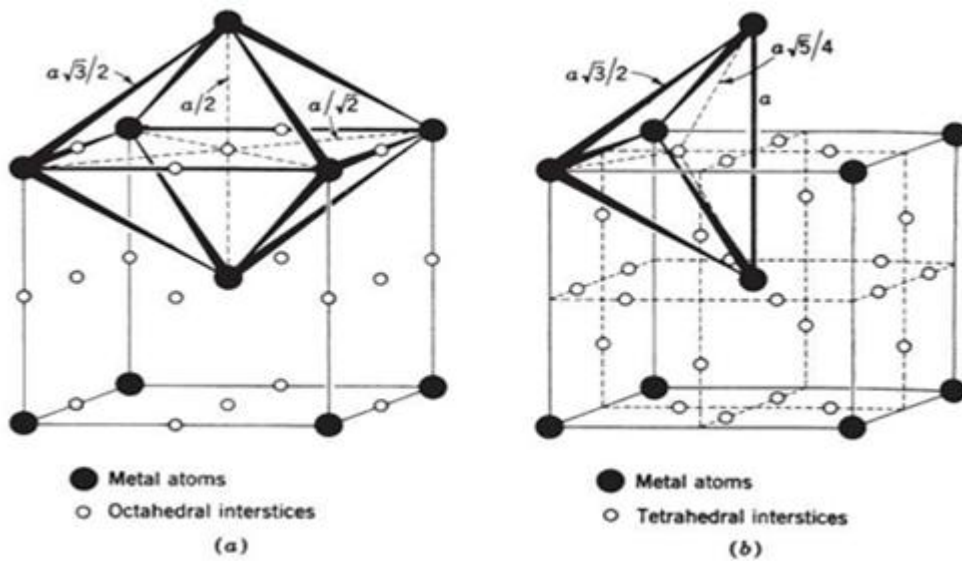


Figure 3.5. (a) Octahedral and (b) tetrahedral interstitial voids in bcc structure [34].

Table 3.1. Solubility of carbon and nitrogen in γ -iron and α -iron [33].

		Temperature (°C)	Solubility	
			wt. %	at. %
C	γ	1150	2,04	8,8
		723	0,8	3,6
	α	723	0,02	0,095
		20	< 0,00005	<0,00012
N	γ	650	2,8	10,3
		590	2,35	8,75
	α	590	0,10	0,40
		20	< 0,0001	<0,0004

3.6. DIFFUSIVITY OF CARBON AND NITROGEN IN FERRIT AND AUSTENITE

The internal friction technique is used to determine diffusion ratios in carbon and nitrogen. The temperature dependence of the diffusivity in the ferrite is the standard exponential relationship as follows;

$$D_C = 6.2 \times 10^{-3} \exp\left(-\frac{Q}{RT}\right) \text{ cm}^2 \text{ s}^{-1} (Q = 80 \text{ kJ mol}^{-1}), \quad \dots\dots\dots(3.1)$$

$$D_N = 3.0 \times 10^{-3} \exp\left(-\frac{Q}{RT}\right) \text{ cm}^2 \text{ s}^{-1} (Q = 76 \text{ kJ mol}^{-1}), \quad \dots\dots\dots(3.2)$$

Here D_C and D_N are the nitrogen and carbon diffusion coefficients and Q is the activation energy. Dependence of D_C and D_N on temperature is shown graphically in Figure 3.6. As can be seen the diffusivity of carbon is slower than nitrogen below 200°C.

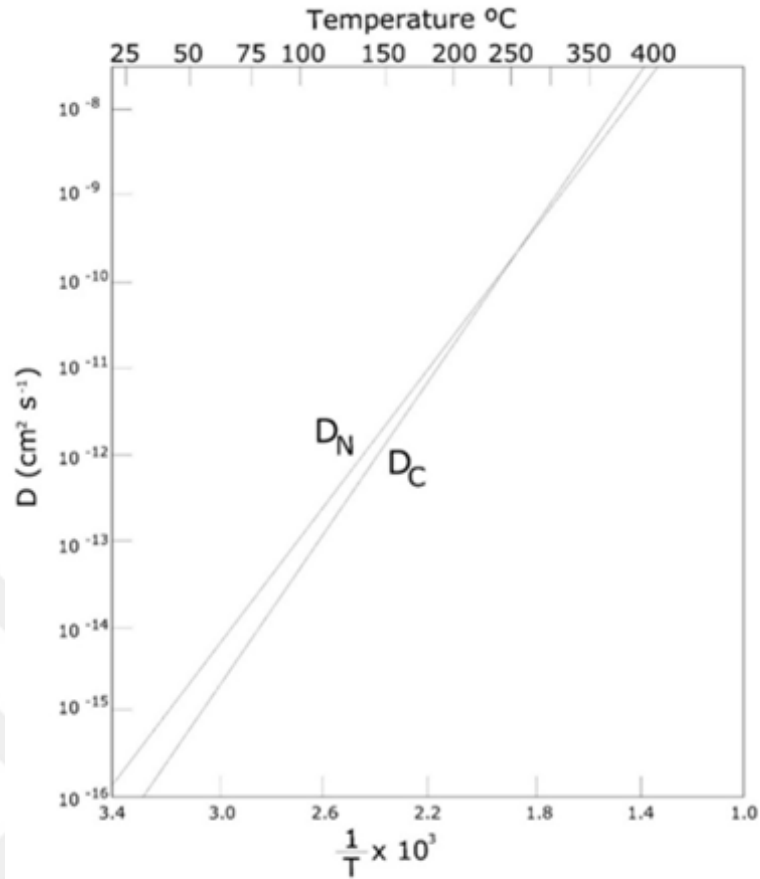


Figure 3.6. Temperature dependence of nitrogen (D_N) and carbon (D_C) diffusion coefficients in α -iron (T is the absolute temperature) [35].

Table 3.2 also shows the diffusion of substitutional and interstitial atoms in α -iron and γ -iron. As can be seen that for a given temperature diffusion of both substitutional and interstitial atoms occurs much faster in ferrite than in austenite.

The reason is that the γ -iron has a tightly packed structure, and loosely packed α -iron reacts more readily to thermal activation and allows easier passage through cavities and associated dissolved atoms. In any case, the activation energy Q is less for a particular element radiating in α -iron; more than for the element that propagates in γ -iron.

Table 3.2. Diffusivity of substitutional and interstitial elements in γ -iron and α -iron.

Solvent	Solute	Activation energy, Q (kJ mol ⁻¹)	Frequency factor, Do (cm ² s ⁻¹)	Diffusion coefficient, D910°C (cm ² s ⁻¹)	Temperature range (°C)
γ -iron	C	135	0,15	1,5x10 ⁻⁷	900-1050
	Fe	269	0,18	2,2x10 ⁻¹³	1060-1390
	Co	364	3,0x10 ²	24,0x10 ⁻¹² (at 1050 C)	1050-1250
	Cr	405	1,8x10 ⁴	58,0x10 ⁻¹² (at 1050 C)	1050-1250
	Cu	253	3	15,0x10 ⁻¹¹	800-1200
	Ni	280	0,77	7,7x10 ⁻¹³	930-1050
	P	293	28,3	3,6x10 ⁻¹²	1280-1350
	S	202	1,35	1,5x10 ⁻⁹	1200-1350
	W	376	1,0x10 ³	12,0x10 ⁻¹² (at 1050 C)	1050-1250
α -iron	C	80	6,2x10 ⁻³	1,8x10 ⁻⁶	
	Fe	76	3,0x10 ⁻³	1,3x10 ⁻⁶	
	Co	240	0,5		700-750
	Cr	226	0,2	2,1x10 ⁻¹¹	700-790
	Cu	343	3,0x10 ⁴		
	Ni	358	9,7	3,7x10 ⁻¹¹	700-900
	P	230	2,9	2,0x10 ⁻¹⁰	860-900
	W	293	3,8x10 ²		

3.7. EFFECT OF ALLOYING ELEMENTS ON STRAIN AGEING

The elements responsible for strain-ageing in steels can be grouped into two groups. In the first group, there are elements such as carbon and nitrogen that can move as fast as the cage structure and interact with dislocations. These elements are evaluated according to resolutions diffusion coefficients and the interaction energy with dislocations. Elements in the second group are elements such as Ti, Nb, Zr and Al that affect the behavior of mobility in the process in the first group and the free content ratios in the cage, which does not affect the ageing process, directly. The first group of elements such as carbon and nitrogen are mainly solved as interstitial atoms in steel. Carbon and nitrogen have important effects on ageing behavior because their

diffusion kinetics is a powerful function of temperature. Therefore, resistance to ageing in steels can be increased by the attachment of elements such as aluminum to nitrogen, and by carbon forming stable carbides. As the amount of dissolution in the form of interstitial atoms in the ferrite structure of nitrogen and carbon increases, the resistance to ageing decreases. In addition to the elements in these two groups, elements such as Mn, Mo and Cr substitutionally soluble in ferritic or austenitic structures and can form interstitial-substitutional couples can be evaluated in another group. These interstitial-substitutional couples can considerably lower the mobility of carbon and nitrogen [36].

The variation of the ageing resistance according to the content ratios of the elements, which are considered to be directly or indirectly affected by ageing, in the unalloyed and low carbon EN 10025-1 S235JR steel is diagrammatically associated with Figure 3.7. Aluminum content was considered when the nitrogen content relationship was examined, and the relationship between the aluminum content and the ageing tendency is schematically shown in Figure 3.7 (a). However, no linear relationship was observed between the aluminum content and the tendency to ageing [37].

Figure 3.7 (b) examines the effect of carbon associations on ageing resistance. As the amount of carbon increases, the tendency of strain-ageing is increasing (ageing resistance is decreasing). But the relationship is not as clear and linear as the associations made on the elemental basis. It has been seen that ageing can be controlled with temperature by controlling the amount of carbides and nitrides deposited and the amount of the remaining interstitial atoms during the rolling of sheets in aluminum-quenched steels. If stoichiometric ratios are established, there are also methods in which the free nitrogen can be attached by the addition of boron, thereby increasing the ageing resistance [37].

In Figure 3.7 (c), the increase in the tendency to ageing with the increase of nitrogen content is observed. Reducing the amount of dissolved carbon or nitrogen depends on the controlled precipitation of the carbide and nitrides precipitates in the ferrite grains. Hayashida et al. [36] observed that the MnS particles larger than 0.05 μm were the preferred points for Fe_3C precipitation in the rolling process performed at

1150°C. When nitrogen and carbon contents were compared, a more significant relationship was observed between ageing tendency and nitrogen content than carbon.

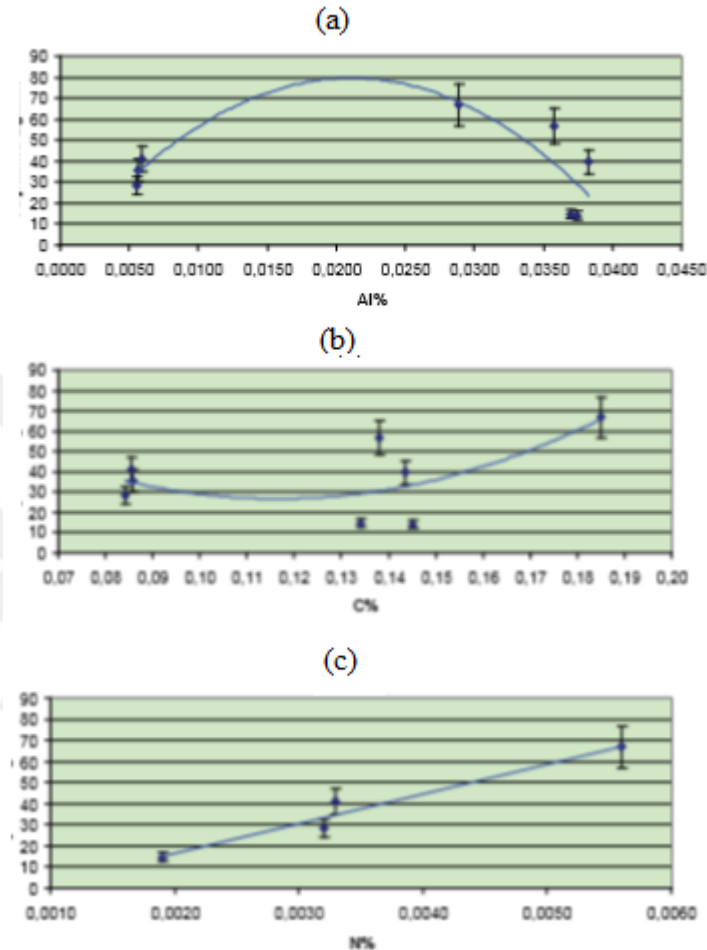


Figure 3.7. Effect of ageing tendency of alloying elements [37].

It has also been attempted in the literature to increase the content of manganese to reduce the amount of free carbon during annealing [38]. The reason of this is, the cementite phase of manganese becomes more stable. The cementite phase forms nucleation areas for stable carbide precipitation. Mangan also reduces the average carbide size in low carbon steels [39]. Abe and colleagues [40] observed that manganese and carbon atoms tend to form manganese-carbon precipitates in ferrite and that these precipitates dissolve in the ageing process and cause carbide formations, based on electrical resistance measurements during annealing in iron-manganese-carbon alloys. In another study, ageing behaviors of aluminum-quenched

steels due to manganese, boron and carbon contents were investigated [41]. In these studies, it can be said that manganese content is effective on carbide precipitation in low carbon steels, and annealing processes based on these mechanisms can produce high strength steels with aging resistance.

3.8. STRAIN AGEING IN AUSTENITIC STAINLESS STEELS

Strain ageing in austenitic stainless steel appears to occur in the austenite phase. In order to understand the reason for the ductile improvement under some intensified conditions, the work-hardening rate was determined and found to be very different between the older and non-elderly structures. It was found that the activation energy of strain ageing in the austenitic stainless steel process was nearly as equal to the activation energy of diffusion of carbon and nitrogen in the austenite phase. A mechanism similar to the Suzuki effect was recommended as the primary mechanism of strain ageing in austenitic stainless steels.

Strain ageing is the interaction between moving dislocations and solute atoms during plastic deformation and is believed to reduce the ductility of the material. However, in this test, some alloys showed an increase in the ductility stage during the strain ageing of austenitic stainless steel. One possible explanation was the formation of twins in austenitic stainless steel. In the austenitic stainless steel strain-ageing process, twisting occurs in some materials when dislocation motion ceases or is limited by solute atoms. This was also known to be; the formation of each twin that lead to some sheared strain.

The strain aging of austenitic stainless steel was studied at a temperature of 550°C with constant strain rates of 2×10^{-2} , 2×10^{-5} , and $2 \times 10^{-5} \text{ sec}^{-2}$. At these strain rates, a characteristic measure of jagged plastic flow was observed over the temperature range of 500°C to 700°C. The strain rate sensitivity of the material was found to be negative in this area. A study of the fracture surface of a tensile specimen by a scanning electron microscope showed that the ductility of the region is reduced. The ultimate drawing ratio of sheet metal is an indicator of deep drawing [42].

CHAPTER 4

STRAIN AGEING IN WELDS

4.1. INTRODUCTION

It has been found that strain ageing causes harmful effects on austenitic stainless steel. Many studies have been carried out on the effect of different parameters on the strain ageing properties of the steels. Generally, strain ageing occurs at about 25, 100, 200, 300, 400, 500, 600, 700 and 800°C in 316L of austenitic stainless steel. These steels are used in structural applications where welding is used as a means of joining structural elements. Due to welding, steel melts in the joint. There is a region away from this fusion zone, where the temperature passes over the temperature range above the steel region in the heat affected zone. The welded materials may be deformed, for example: bent plates before welding in the manufacture of ships. These deformed materials passing through the temperature range above may be subject to strain ageing.

4.2. FUSION WELDING OVERVIEW

4.2.1. Fusion Welding Processes

The fusion welding is a joining process that utilizes fusion of the base metal to weld. The three main fusion welding types are as follows:

- **Gas welding:**
 - i. Oxyacetylene welding (OAW)
- **Arc welding:**
 - i. Shielded metal arc welding (SMAW)
 - ii. Gas–tungsten arc welding (GTAW)
 - iii. Plasma arc welding (PAW)

- iv. Gas-metal arc welding (GMAW)
- v. Flux-cored arc welding (FCAW)
- vi. Submerged arc welding (SAW)
- vii. Electroslag welding (ESW)
- **High-energy beam welding:**
 - i. Electron beam welding (EBW)
 - ii. Laser beam welding (LBW)

Since there is no arc in the electroslag welding process, it is not exactly an arc welding process. Arc welding processes is grouped for convenience of discussion [13].

4.2.2. Power Density of Heat Source

The heat sources for the gas, high-energy beam and arc welding processes are a gas flame, a high-energy beam and an electric arc, respectively. The power density increases from a gas flame to a high-energy beam and to an electric arc. As shown in Figure 4.1, the power density of the heat source increases with the decreases in the heat input to the workpiece required for welding. A part of the workpiece material that is exposed to gas is heated so slowly that before it reaches any melting point, a large amount of heat is transferred to the mass of the workpiece. Conversely, the same material, which is focused electronically or laser beam, can be evaporated to form an immediate and deep key hole, and even welded without too much heat being transferred to the mass of the workpiece. [43].

Thus, the advantages of increasing the power density of the heat source as shown in Figure 4.1 are better welding quality, with deeper source penetration, less damage to the workpiece and higher welding speeds. Figure 4.2 shows that as the heat input per unit length of the weld per unit thickness of the samples decreases, the weld strength (aluminum alloys) increases [27].

Angular distortion is much bigger in GTAW (2) than in EBW as shown in Figure 4.3.a. Unfortunately, the cost of electron and laser beam welding machines is very high, as shown in Figure 4.3b [44].

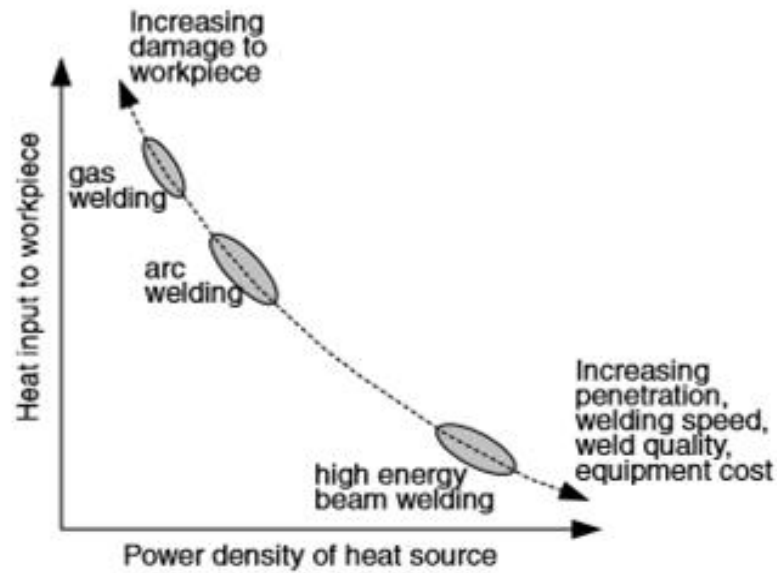


Figure 4.1 Variation of heat input to the workpiece with power density of the heat source [44].

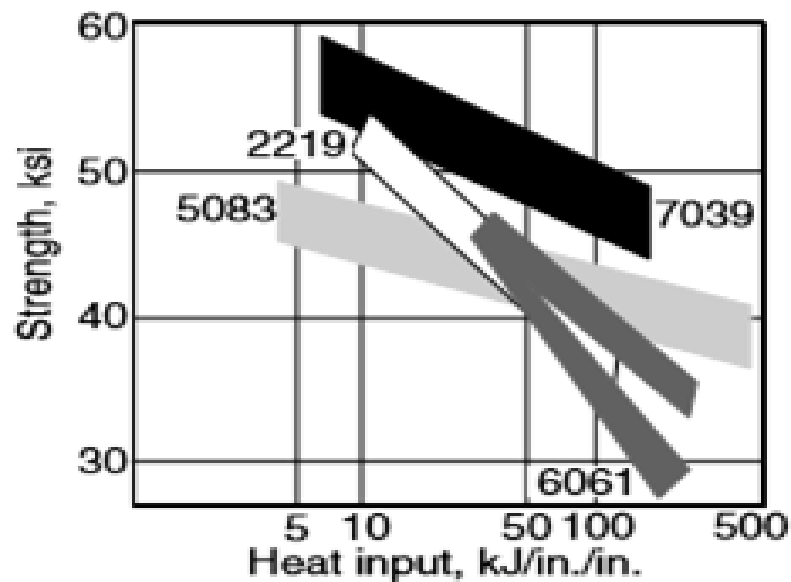


Figure 4.2 Variation of weld strength with heat input per unit length of weld per unit thickness of samples [44].

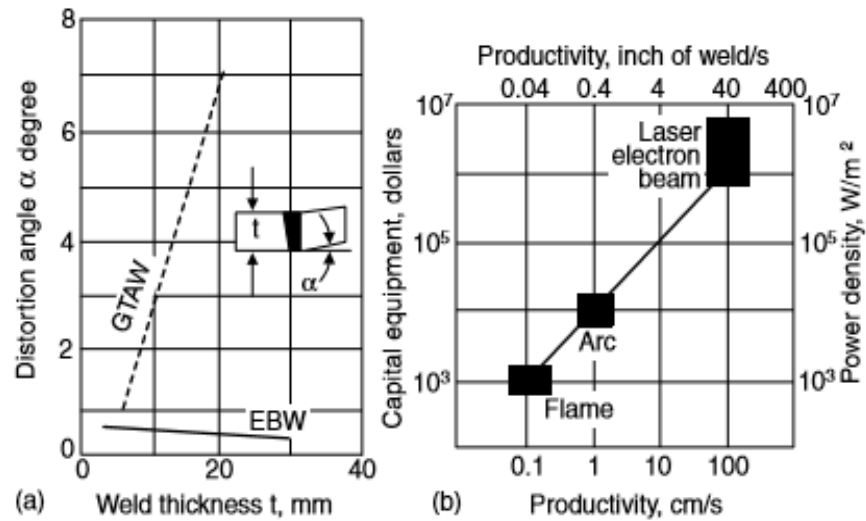


Figure 4.3 Comparisons between welding processes: (a) angular distortion, (b) capital equipment cost [44].

4.2.3. Welding Processes and Materials

Table 4.1 summarizes the fusion welding processes recommended for carbon steels, stainless steels, low-alloy steels, cast irons, aluminum alloys and nickel-base alloys [28]. For one example, GMAW can be used for all the materials of almost all thickness ranges while GTAW is mostly for thinner workpieces. For another example, any arc welding process that requires the use of a flux, such as SMAW, SAW, ESW and FCAW is not applicable to aluminum alloys [45].

Table 4.1. Overview of welding processes [45].

Material	Thickness	SMAW	SAW	GMAW	FCAW	GTAW	PAW	ESW	OFW	EBW	LBW
Carbon steels	S	X	X	X		X			X	X	X
	I	X	X	X	X	X			X	X	X
	M	X	X	X	X				X	X	X
	T	X	X	X	X			X	X	X	
Low-alloy steels	S	X	X	X		X			X	X	X
	I	X	X	X	X	X				X	X
	M	X	X	X	X					X	X
	T	X	X	X	X			X		X	
Stainless steels	S	X	X	X		X	X		X	X	X
	I	X	X	X	X	X	X			X	X
	M	X	X	X	X		X			X	X
	T	X	X	X	X			X		X	
Cast iron	I	X							X		
	M	X	X	X	X				X		
	T	X	X	X	X				X		
Nickel and alloys	S	X		X		X	X		X	X	X
	I	X	X	X		X	X			X	X
	M	X	X	X			X			X	X
	T	X		X				X		X	
Alumminum and alloys	S			X		X	X		X	X	X
	I			X		X				X	X
	M			X		X				X	
	T			X						X	

4.2.4. Types of Joints and Welding Positions

Figure 4.4 shows the basic welding connection designs in the fusion welding: butt, T, edge and corner connections and lap. The two junctions between the face and the workpiece surface are called the toes, the surface of the weld is called the face, and the portion of the weld beyond the workpiece surface is called the reinforcement [46].

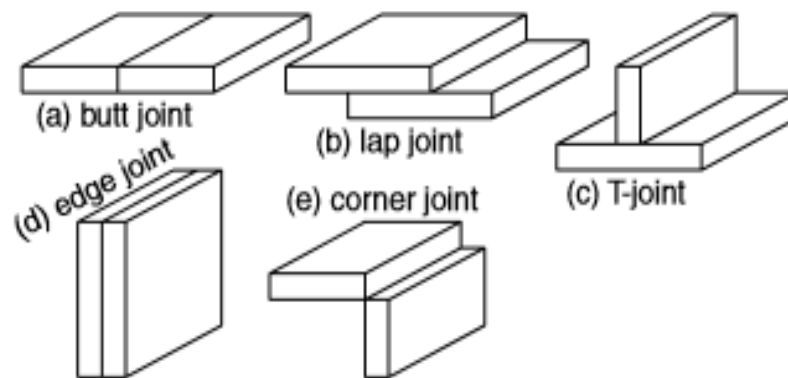


Figure 4.4 Five basic types of weld joint designs.

4.3. STRESS AND STRAINS AROUND WELDS

All welding processes involve a heat source which is the main cause of melting. The formation of an integral joint should be formed by subsequent solidification. Most of the heat is spread to adjacent solid regions from the fusion region. As a result, these regions experience a significant degree of heating and cooling cycle relative to distance from the fusion boundary. The farther away from the fusion boundary known as the heat affected zone, the lower the maximum temperature and heating rate. This also defines the temperature range at which the austenitic phase is transformed by solid state transformation. At the time of welding, very high local heating produces expansion which is prevented by the restraint of that part of the piece which remains cold. This results in extensive compressive plastic deformation in the heated zone. On cooling, restraint by the cold part creates residual tensile stresses in the longitudinal and transverse direction of the welded joint [27, 47].

It is not surprising therefore that the metallurgical effects of the weld thermal cycle are complex and may in some instances result in an unfavourable change in the properties of the material [48]. Consider a butt weld as shown in Figure 4.5, a section of material in the vicinity of the weld experiences different rates of expansion and contraction compared with other sections around it. In tracing the behaviour of this section during a weld thermal cycle, changes in the section's mechanical properties as a function of temperature, e.g. elastic modulus (E), yield strength, Poisson's ratio,

etc., have also to be taken into account. These properties can change drastically at temperatures as low as 500-600°C in steel [49].

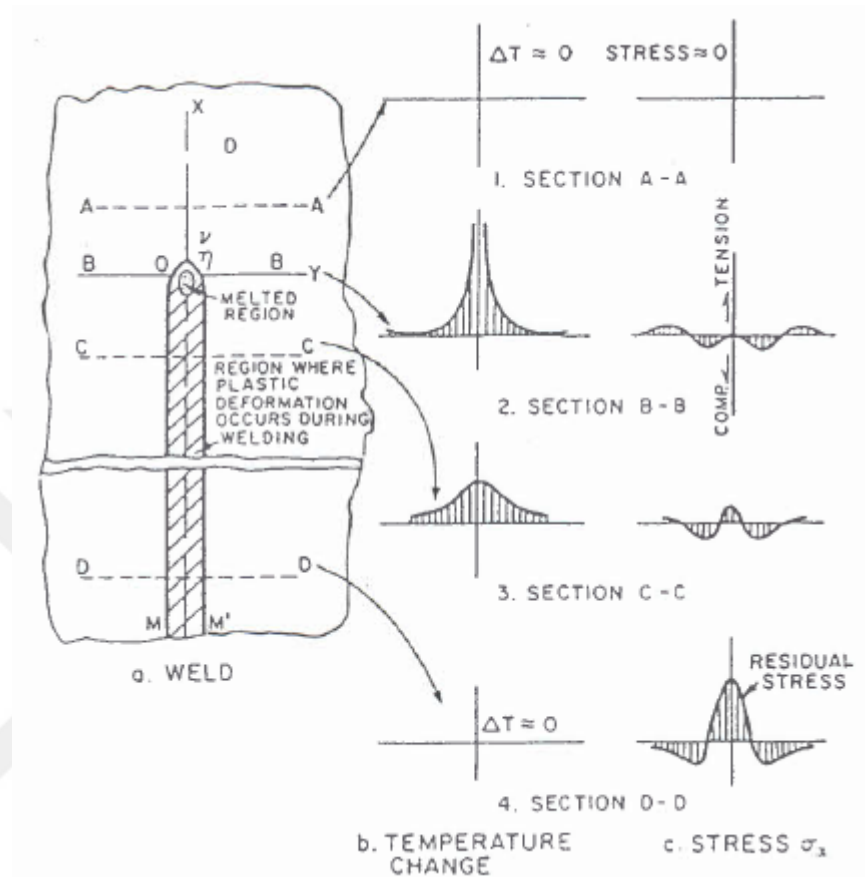


Figure 4.5 Schematic representation of changes of temperature and stress during welding [49].

A very trying practical problem, which frequently arises in fabricating weldments, is to be certain that the finished article will conform to the required dimensions and will have all dimensions in proper proportion. Inaccuracies in dimensions or form easily develop because of the unavoidable heat effects which accompany most welding processes. Although these conditions are unavoidable, they can be minimised and in many cases, one condition can be used to counteract another. If a few basic facts concerning the behaviour of metal during heating and cooling while under restraint are understood, a little engineering skill can accomplish a great deal in securing weldment accuracy [50].

4.4. EFFECT OF WELDING ON THE STRUCTURE AND PROPERTIES OF MATERIALS

The weld metal structure of the weld joint is strongly influenced by the chemical composition of the filler metal and the process heat input. In most cases, the higher heat input results in a slower cooling rate, which leads to the coarse grains in the weld metal. Moreover, the lower heat input results in a rapid cooling rate, leading to a fine microstructure. In refining the weld metal microstructure, the natural properties of the process play an important role, although it produces finer grains of higher heat input and lower heat input. Although the mechanical effect of the refined microstructure process and the rapid cooling rate (i.e. lower heat input) are combined, it is possible to form the upper bainite containing a small amount of acicular ferrite.

SMAW and GMAW weld metal undergoes a phase change due to higher heat input. The parent material microstructure (ferrite and a small amount of pearlite) is converted to acicular ferrite by heat input and chemical composition with a small amount of retained austenite and martensite. Variations in the chemical composition of the parent and filler metal leads to the thermal variations of the parent metal and weld metal and as well as the solidification of the weld metal. A slow cooling rate may reduce the interfacial energy between austenite and ferrite to form acicular ferrite. Acicular ferrite has the potential to combine high toughness and strength tissue stability in acicular ferrite is higher at higher temperatures than bainite [13].

Due to the high heat input and strength of the GMAW process, coarse grains in the weld metal are formed. The GMAW process cools relatively slowly, resulting in the formation of martensite and acicular ferrite with retained austenite in the weld metal area. Moreover, the addition of filler metal also plays a crucial role in the weld metal structure. It is impossible to produce a uniform weld metal by the fusion welding process. Strong weld metal, weld metal chemical composition and microstructure control weld performance. Very strong carbide / nitride forming elements such as Nb, Ti, V have only very limited solubility in ferrite and austenite, and the precipitates generally function as fine dispersions of carbides, nitrides and / or

carbonitrides, and due to precipitation hardening. This may be the reason that the yield strength of all joints is increased as compared to the yield strength of the base metal. The yield point extension is due to the interaction of solute atoms and dislocations [46].

A considerable increase in molybdenum and nickel may lead to the formation of ferrite with martensite and retained austenite due to the addition of filler metals and high heat input. This shows that the strength of the welded joint is reduced by 7% compared to the parent metal due to an increase in the original austenite grain size in the reheated and isothermally held samples. Ductility and microstructure and chemical composition changes more complicated. In general, in addition to the grain size, all the factors that increase the strength will reduce the ductility [46]

4.5. HEAT AFFECTED ZONE (HAZ)

Identification of the heat affected zone of the parent metal is difficult. It will depend upon weld time, the temperature of the weld pool, welding methods, thermal conductivity and dimensions of the parent metal. Welded joints produced in metals such as aluminum and copper with high thermal conductivity will have a larger area exposed to heat than a flat carbon steel with a lower thermal conductivity. According to the gas source, the metal arc welding provides a more concentrated heating effect. Because of this, the welding operation is faster as the heat energy output is larger than the arc welding, For this reason, when the gas welding is made from the same welding, the arc source will be narrower than the heat welding region.

In the mild steel plate, the heat affected zone exhibits various structures. They are heated to an extent that the temperature above the room temperature rises too high above the upper critical temperature for steel that is overheated for parts adjacent to the weld pool. These are shown in Figure 4.6 for a single run oxyacetylene weld and a single run metallic arc welding [51].

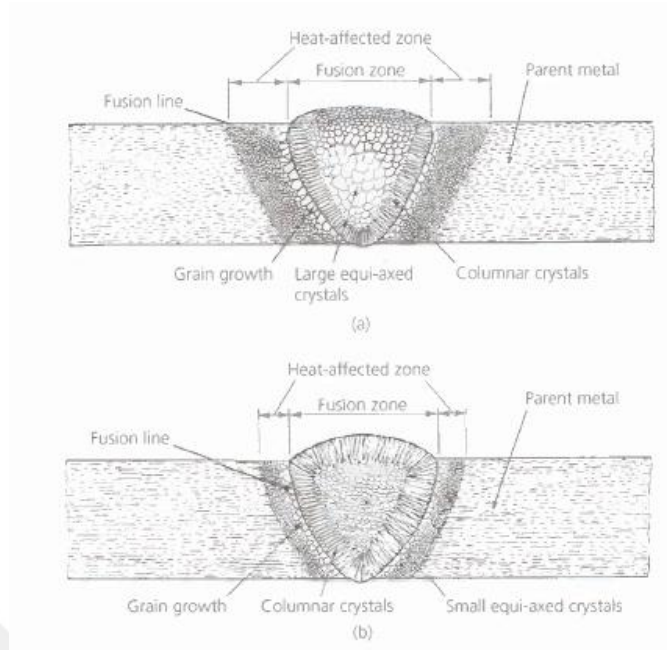


Figure 4.6 Microstructure of single run welds in mild steel: (a) oxyacetylene welds and (b) metallic arc weld [51].

Figure 4.7 shows the weld microstructure of a work-hardened 304 stainless steel. Before workhardening, the microstructure of the same material is also included in the comparison (Figure 4.7a). Recrystallization (Figure 4.7d) and grain growth (Figure 4.7e) are evident in HAZ. Figure 4.8 shows the grain growth of the molybdenum weld in HAZ [52]. Severe HAZ grain growth may cause coarse grains in the fusion region due to epitaxial growth. Fracture toughness is usually poor with coarse particles in fusion zone and the HAZ.

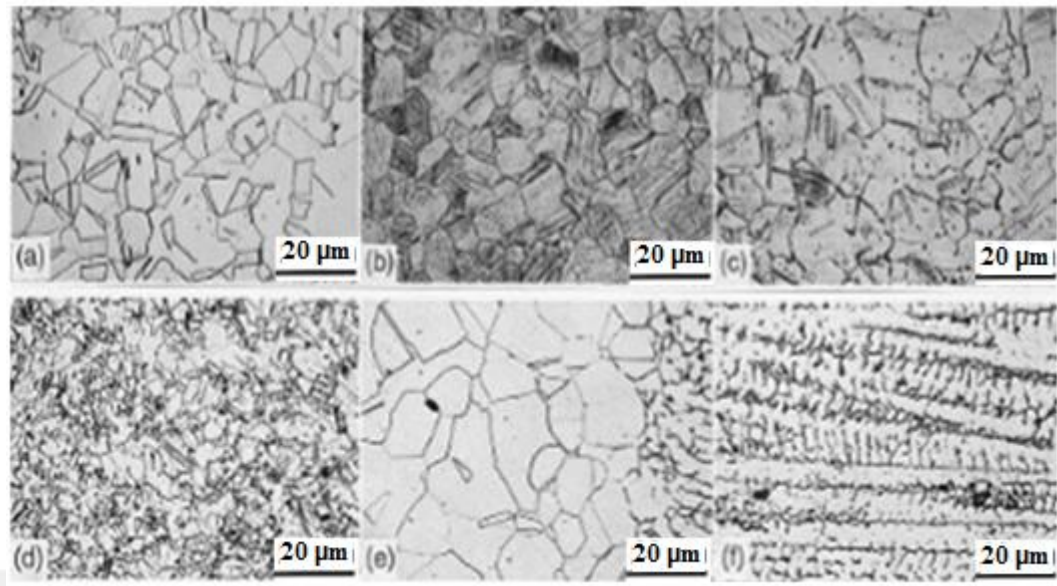


Figure 4.7 Microstructure across the weld of a work-hardened 304 stainless steel: (a) before work hardening; (b) base metal; (c) carbide precipitation at grain boundaries; (d) recrystallization; (e) grain growth next to fusion boundary; (f) fusion zone [52].

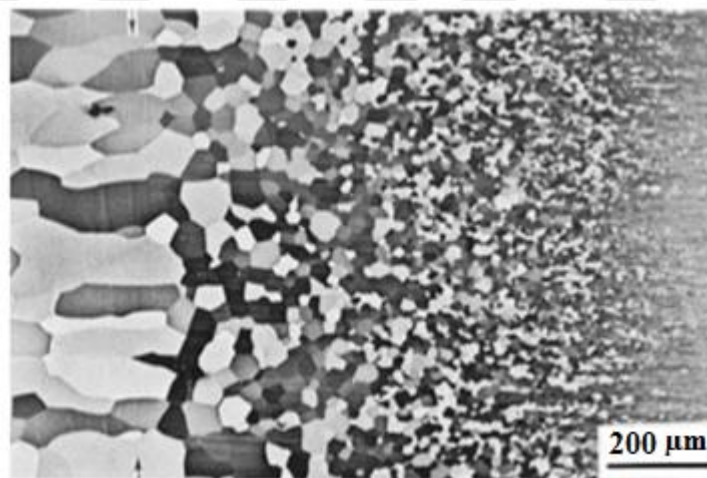


Figure 4.8 Grain growth in electron beam weld of molybdenum, arrows indicating fusion boundary. Reprinted from Wadsworth et al. [52].

CHAPTER 5

EXPERIMENTAL MATERIALS AND METHODS

5.1. INTRODUCTION

This chapter contains a description of the experimental materials and methods used to investigate dynamic strain ageing in a 316L austenitic stainless steel under as-received, as-welded and weld metal conditions. Welding process was conducted using the MIG welding technique. Tensile test was performed on as-received sample, as-welded sample and weld metal at $1 \times 10^{-3} \text{ s}^{-1}$ strain rate for the temperatures of 25-800°C to determine the influence of dynamic strain ageing on the microstructure and strength of investigated material.

The chapter begins with an explanation of steel and welding wire composition (section 5.2). Section 5.3 describe details of the welding procedure. This is followed by a description of the experimental methods used to produce tensile test specimens from as-received samples, as-welded samples and weld metal (section 5.4). An experimental method used for dynamic strain ageing and hardness measurement is also described in sections 5.5 and 5.6 respectively. Preparation and analyses of specimens for microscopical examination was carried out in sections 5.7 and 5.8 respectively.

5.2. MATERIALS

In this study, 316L austenitic stainless steel was selected to investigate the dynamic strain ageing behaviour. AS MIG 316LSi welding wire in diameter of 1.2 mm was also used in the present work. The chemical composition of welding wire and the steel used for joining is given in Table 5.1. The steel was supplied in plates with dimensions of $8 \times 150 \times 350 \text{ mm}$.

As shown in Figure 5.1, grooves with V-groove configuration at 30° was made on the edge of 316L austenitic stainless steel for welding.

Table 5.1. Chemical compositions of 316L austenitic stainless steel and welding wire (wt. %).

Materials	C	Si	Mn	Ni	Cr	Mo	N	S	P	Fe
316L	0.024	0.38	1.30	10.10	16.57	2.03	0.041	0.004	0.029	Bal.
Welding wire	0.03	0.85	1.70	12.5	18.5	2.75	-	0.003	0.0025	Bal.

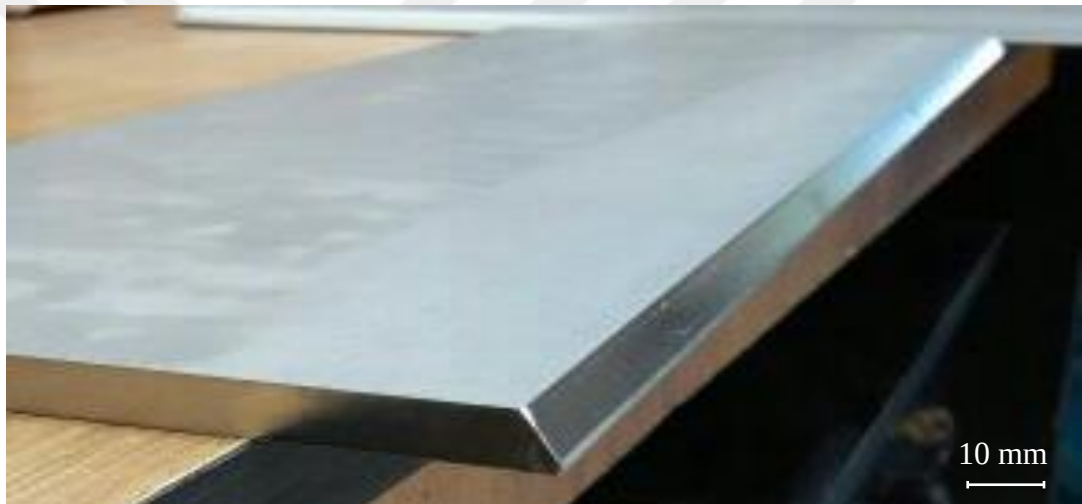


Figure 5.1. Grooves which was made on the edge of 316L austenitic stainless steel with V-groove configuration at 30° for welding.

5.3. WELDING OF MATERIALS

The steel plates, except those in the as-received conditions were welded longitudinally using Fanuc Robotic Lincoln Electric brand MIG welding machine by a single pass through a V-groove configuration at 30°. Welding was performed under argon gas by using the welding parameters as indicated in Table 5.2. Fig. 5.2 shows welded plates.

Table 5.2. The welding parameters used in the gas metal arc welding (GMAW).

Current (A)	Voltage (V)	Welding Speed (cm/min)	Gas Flow Rate (lt/min)	Heat Input (kj/mm)
240	23	35	14	0.95

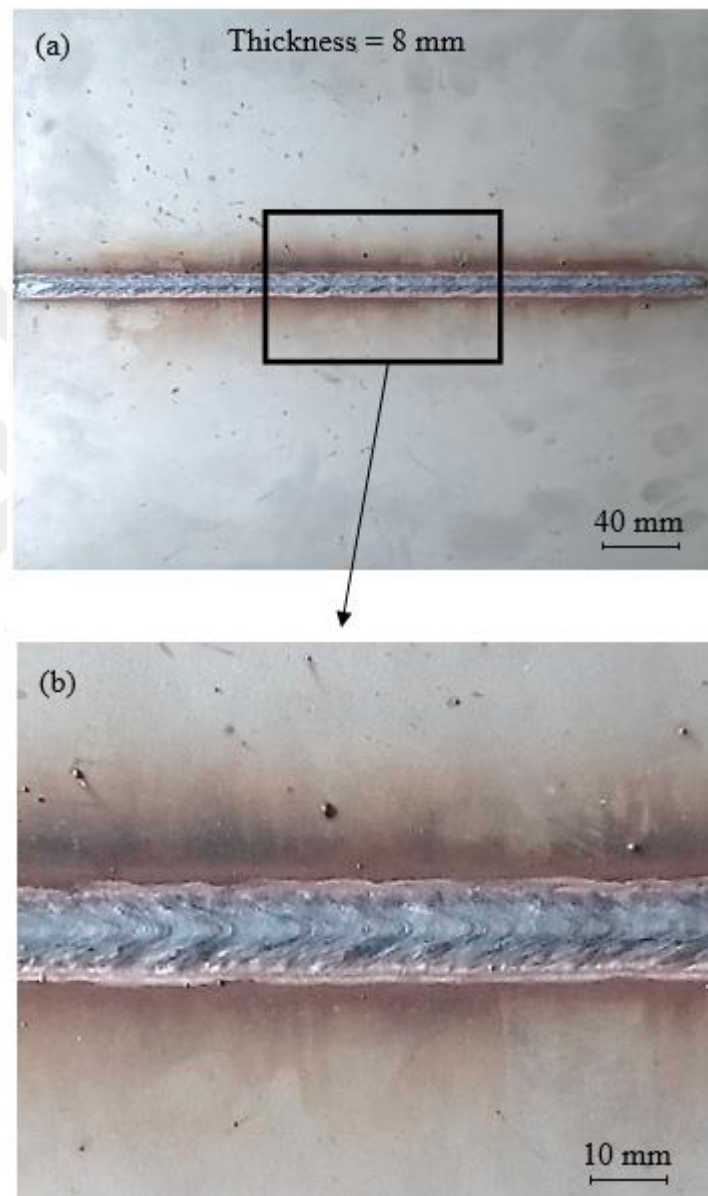


Figure 5.2. 316L austenitic stainless steel welded by Fanuc Robotic Lincoln Electric brand MIG welding machine: (a) full range profiles and (b) sections in large scala.

5.4. PRODUCTION OF TENSILE TEST SPECIMENS FROM AS-RECEIVED AND WELDED MATERIALS

Specimens were cut from the as-received, as-welded and weld metal blanks for evaluation of the tensile testing. Welded sheets were cut vertically to the welding direction as shown in Fig. 5.3 to obtain as-welded specimens. The weld region was in the middle of the specimens for welded joints.

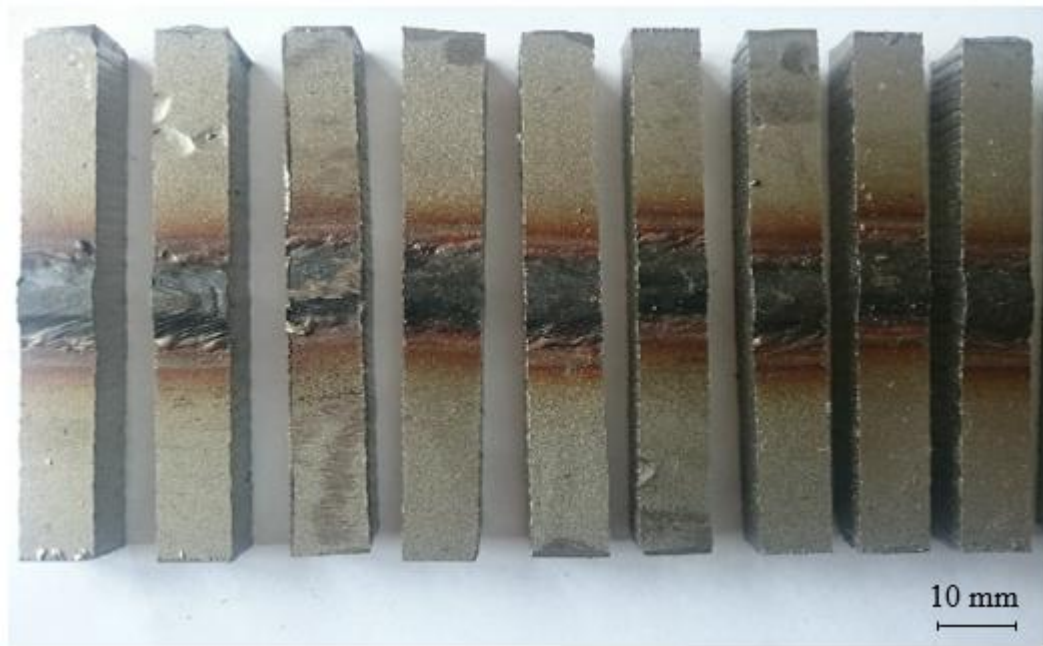


Figure 5.3. Sampling of as-welded samples cut vertical to the welding direction.

Specimens for tensile test were manufactured from the as-received, as-welded and weld metal blanks. Tensile test pieces with gauge length parallel to the rolling direction were manufactured in accordance with the relevant standard E8 (30 mm gauge length and 5 mm diameter) as shown in Figure 5.4.



Figure 5.4. Tensile test specimen used for hot tensile test.

5.5. TENSILE TEST FOR DYNAMIC STRAIN AGEING

Tensile tests for dynamic strain ageing were carried out using a MTS (100kN Servohydraulic Dynamic Tester) in the temperature range of 25-800°C at a strain rate of $1 \times 10^{-3} \text{ s}^{-1}$ for as-received sample, as-welded sample and weld metal. Care was taken to mount the tensile test pieces parallel to the tensile test direction in the grips. After each test, stress and strain diagrams were obtained. From these, elongation (%) work hardening index (n), yield strength (YS, 0.2 %) and ultimate tensile strength (UTS) were derived.

True stress (σ) and true strain (ϵ) and log true stress and log true strain were calculated to obtain the work hardening index (n) up to the maximum strength for the homogeneous plastic deformation zone. The slope of the second relationship defines the workhardening index (n) which gives a quantitative measure of the workhardening characteristic of as-received sample, as-welded sample and weld metal. It can be defined as [53]:

$$\sigma = K\epsilon^n \quad (5.1)$$

$$\log \sigma = \log K + n \log \epsilon \quad (5.2)$$

Where σ is the true stress, ϵ is the true strain, n is the work hardening index, and K is the strength coefficient.

5.6. HARDNESS MEASUREMENT

Vickers hardness measurements of welded joints were carried out at 500 g load and a diamond square-based pyramid, which gives geometrically similar impression under load. A microscope was swung over the square impression, the diagonals of which are measured between knife edges and the readings were taken from a digital counter. 13 hardness measurements were made at 1 mm intervals along the width of the welded samples. Fig. 5.5 shows Shimadzu micro hardness test machine used for the measurement of Vickers hardness.



Figure 5.5. Shimadzu microhardness test machine used for the measurement of Vickers hardness.

5.7. PREPARATION OF SPECIMENS FOR MICROSCOPICAL EXAMINATION

In the present work, scanning electron microscopy (SEM) and optical microscopy were used to characterize steel microstructure under as-received, as-welded and weld metal conditions. Specimens for microscopical examination were obtained from the gauge length of broken tensile pieces where sectioned perpendicular to the tensile

axis. The microscopical examination has been carried out for the as-received, as-welded and weld metal conditions.

Obtained test pieces for metallographic examinations are so small that it is difficult to hold it flat on the silicon carbide papers. Therefore, each piece was mounted in Bakelite which softens at about 150°C, which is usually too low a temperature to cause any structural change in the specimen. The specimens were placed face down at center of piston and then two spoons of Bakelite powder were added in the mould. After, the powder was electrically heated and a pressure up to about 30 kN applied. Afterwards, the mounted specimen was allowed to cool for a fixed periods of time and finally, it was removed from the mould. Figure 5.6 shows a mounting machine used for mounting. Figure 5.7 also reveals a sample which was mounted by mounting machine.



Figure 5.6. Machine used for mounting the specimens.

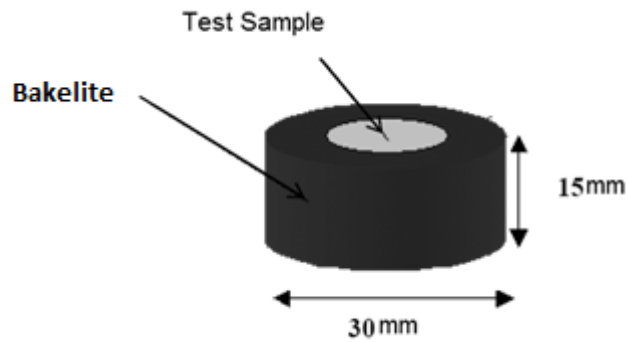


Figure 5.7. A sample which was mounted by mounting machine.

When preparing a specimen for microscopic examination, it had to be produced on a surface that looked perfectly flat and scratch-free when first viewed with a microscope. The samples were wet ground to 1200 mesh using a coarse, medium and finer grinding on a set of silicon carbide paper and then polished to a 1 micron skin using a diamond abrasive compound to remove traces from the grinding process.

The polishing process causes a very thin layer of amorphous metal to be burnished over the surface of the specimen, thus hiding the crystal structure. In order to observe the microstructure of the samples, a mixture of 89% distilled water, 10% oxalic acid and 1% nitric acid solution was used for electrolytic etching, which was done at room temperature and a potential of 15-20 V. Figure 5.8 shows grinding and polishing machine.



Figure 5.8. Machine used for grinding and polishing.

5.8. ANALYSES OF SPECIMENS FOR MICROSCOPICAL EXAMINATION

Microstructure and fracture surface examinations of as-received samples, as-welded samples and weld metal were carried out by means of optic and scanning electron microscopes (SEM) with energy dispersive spectrometry (EDS) analyzer. The optical microscope examination of samples tested at temperatures of 25°C, 200°C, 400°C, 600°C and 800°C was carried out using a Nikon Epiphot microscope capable of magnifications between $\times 50$ and $\times 1000$. Optical micrographs were obtained from the areas of the sample which were characteristic of the whole microstructure. Fig. 5.9 shows Nikon Epiphot type microscope used in the present experimental work.



Figure 5.9. Optical microscope used in the present experimental work.

The fracture surface of broken tensile test specimens which were tensile tested at temperatures of 25°C, 200°C, 400°C, 600°C and 800°C was examined by using Carl Zeiss Ultra Plus Gemini Fesem type SEM. The fracture morphology of as-received sample, as-welded sample and weld metal was determined. Figure 5.10 shows SEM which were used to examine fracture surface of samples under different conditions.



Figure 5.10. Scanning electron microscopy used in the present experimental work.

CHAPTER 6

RESULTS AND DISCUSSIONS

6.1. INTRODUCTION

In this chapter, dynamic strain ageing response of 316L austenitic stainless steel under as-received, as-welded and weld metal conditions was discussed by presenting related results. Tensile test was performed on as-received samples, as-welded samples and weld metal at $1 \times 10^{-3} \text{ s}^{-1}$ strain rate for the temperatures of 25-800°C to determine the influence of dynamic strain ageing on the microstructure and strength of investigated material. The chapter begins with the results and discussion of optical microstructure for as-received samples, as-welded samples and weld metal (section 6.2). Tensile test results of as-received samples, as-welded samples and weld metal are discussed in section 6.3. This is followed by a discussion of the SEM results of as-received samples, as-welded samples and weld metal in section 6.4.

6.2. MICROSTRUCTURE RESULTS AND EVALUATION

6.2.1. Evaluation of Microstructure Results for As-received Samples

Figure 6.1 shows optical microstructure of as-received samples. As can be seen, structures consisted of typical austenite grains with different sizes. The microstructure of the as-received metal was austenitic with a low amount of delta ferrite. Ideally, austenitic stainless steels contain a single phase which is maintained over a wide temperature range.

However, a fully austenitic microstructure is more hot crack sensitive to the microstructure containing a small amount of ferrite [54]. Figure 6.2a-d also shows the microstructure near fracture surface of the as-received samples tested at 200°C, 400°C, 600°C and 800°C respectively for the strain rate of $1 \times 10^{-3} \text{ s}^{-1}$. As is

seen samples tested at different temperatures showed deformation bands and twins in the grains.

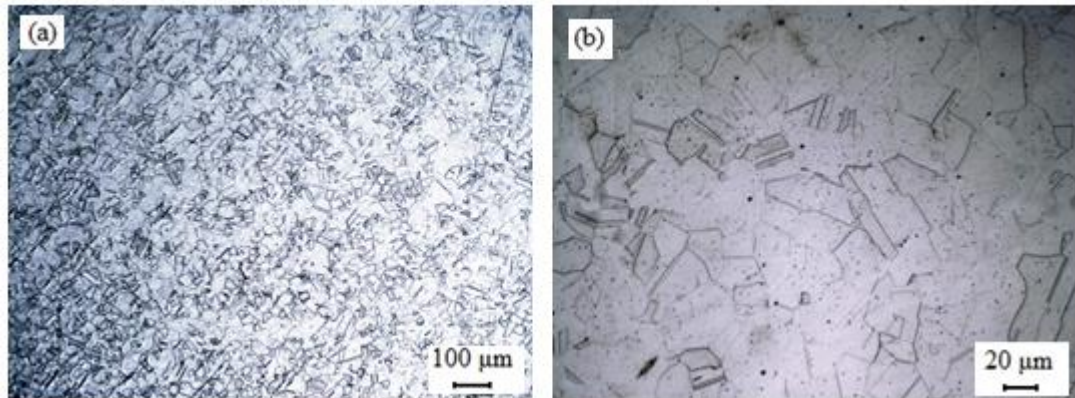


Figure 6.1 Microstructure of 316L austenitic stainless steel: (a) lower magnification and (b) higher magnification.

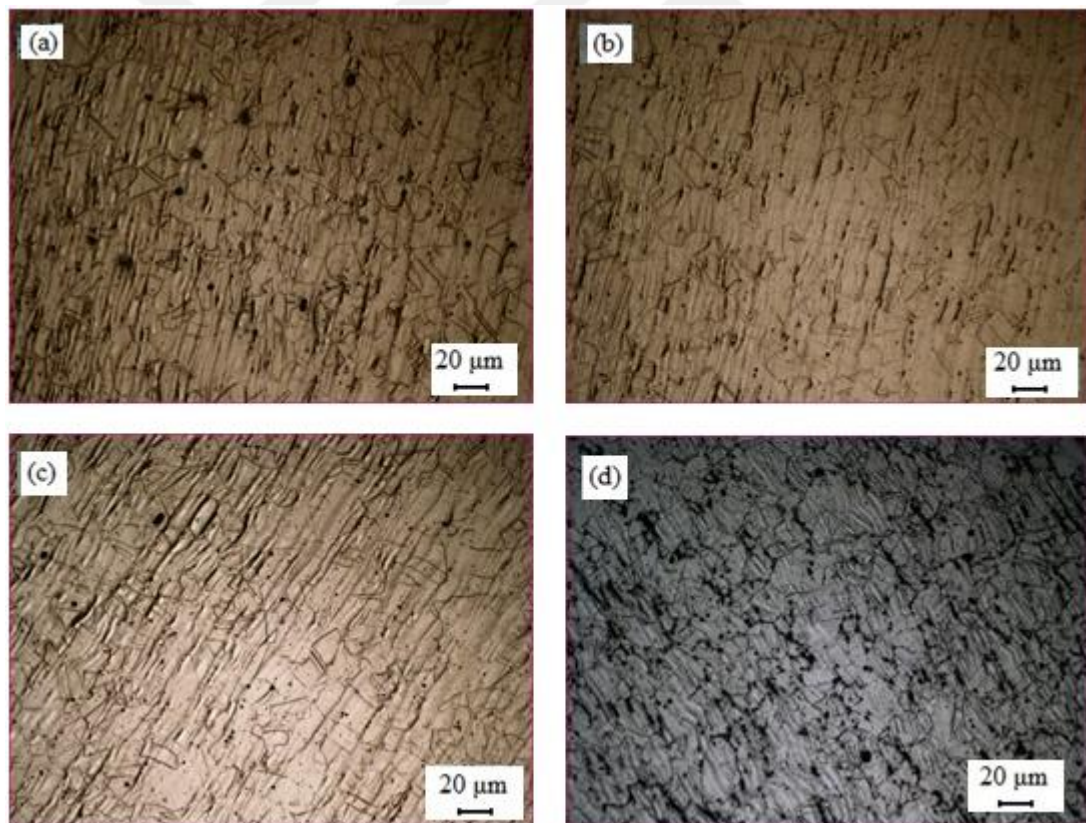


Figure 6.2 Microstructure of 316L austenitic stainless steel tested at (a) 200°C, (b) 400°C, (c) 600°C and (d) 800°C.

6.2.2. Evaluation of Microstructure Results for As-welded Samples

Figure 6.3a–d shows optic microscopy images of the as-welded samples. Three different zones were observed as indicated in Figure 6.3a. The weld metal (Figure 6.3b) showed a solidification microstructure due to the welding thermal cycle [13]. This consisted of skeletal ferrites in an austenitic matrix of large columnar austenite grains. The alignment is along the heat flow direction, which is also the primary dendrite growth direction. The ferrite is located within the cores of the primary and secondary dendrite arms and is the result of the incomplete primary $\delta \rightarrow \gamma$ transformation [54]. HAZ occurred by the peak temperatures and cooling rates showed any austenite grains located beside the fusion lines as seen in Figure 6.3c. As discussed in previous section the microstructure of the as-received metal (Figure 6.3d) was austenitic with a low delta ferrite.

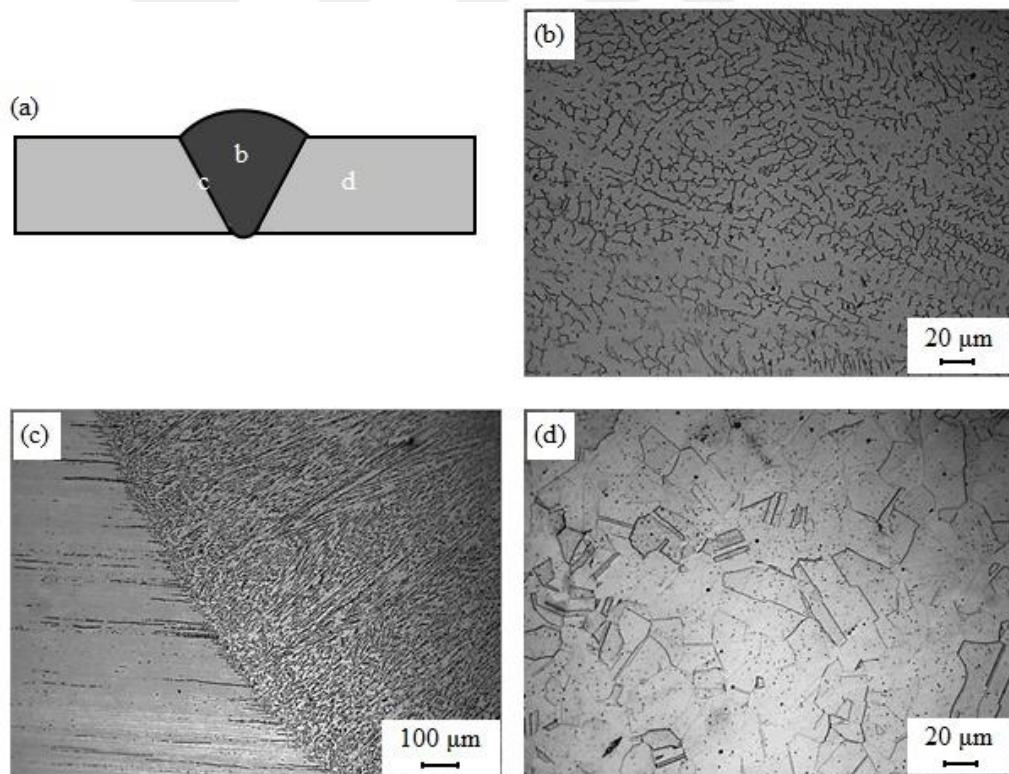


Figure 6.3 Optical micrographs of (a) schematic representation of welded joint obtained by the GMAW process, (b) weld metal, (c) HAZ recrystallized, and (d) base metal.

6.2.3. Evaluation of Microstructure Results for Weld Metal

Figure 6.4 shows weld metal microstructure of 316L austenitic stainless steel tested at 25°C for the strain rate of $1 \times 10^{-3} \text{ s}^{-1}$. During the welding process, a solidification microstructure consisting of skeletal ferrite in large columnar austenite grains was observed in weld metal. The alignment of ferrite is along the heat flow direction due to incomplete primary $\delta \rightarrow \gamma$ transformation [13, 55]. The rapid cooling of the weld pool during welding process causes the formation of ferrite in the weld metal. Ferrite phase in weld metal increased due to rapid cooling after welding which prevented austenite phase transformation. This is consistent with earlier studies [56, 57]. Figure 6.5a-d also shows the microstructure near fracture surface of the weld metal samples tested at 200°C, 400°C, 600°C and 800°C respectively for the strain rate of $1 \times 10^{-3} \text{ s}^{-1}$. As can be seen, samples tested at different temperatures showed similar weld metal microstructure as the samples tested at 25°C consisted of austenite grains with skeletal ferrite.

In many applications, the ability to control the delta-ferrite content of stainless steel weld metal is important. For example, the ferrite number (FN) is often used as an indicator of the resistance to fissuring (hot tearing) that occurs in many of the 300-series stainless steels. Delta ferrite, rich in Cr and Mo, occurred during the welding of austenitic stainless steel is required up to a limit of about 4–5 ferrite number (FN) to prevent micro fissuring or hot tearing of the weld metal [58].

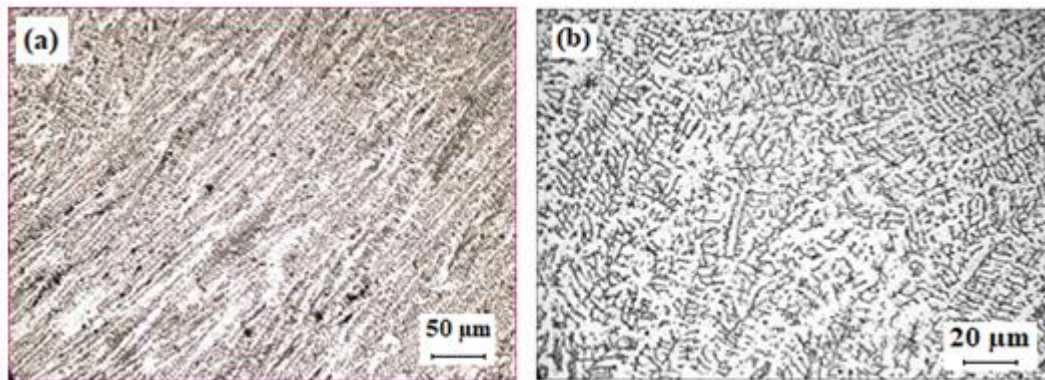


Figure 6.4 Weld metal microstructure of 316L ASS steel tested at 25°C; (a) lower magnification and (b) higher magnification.

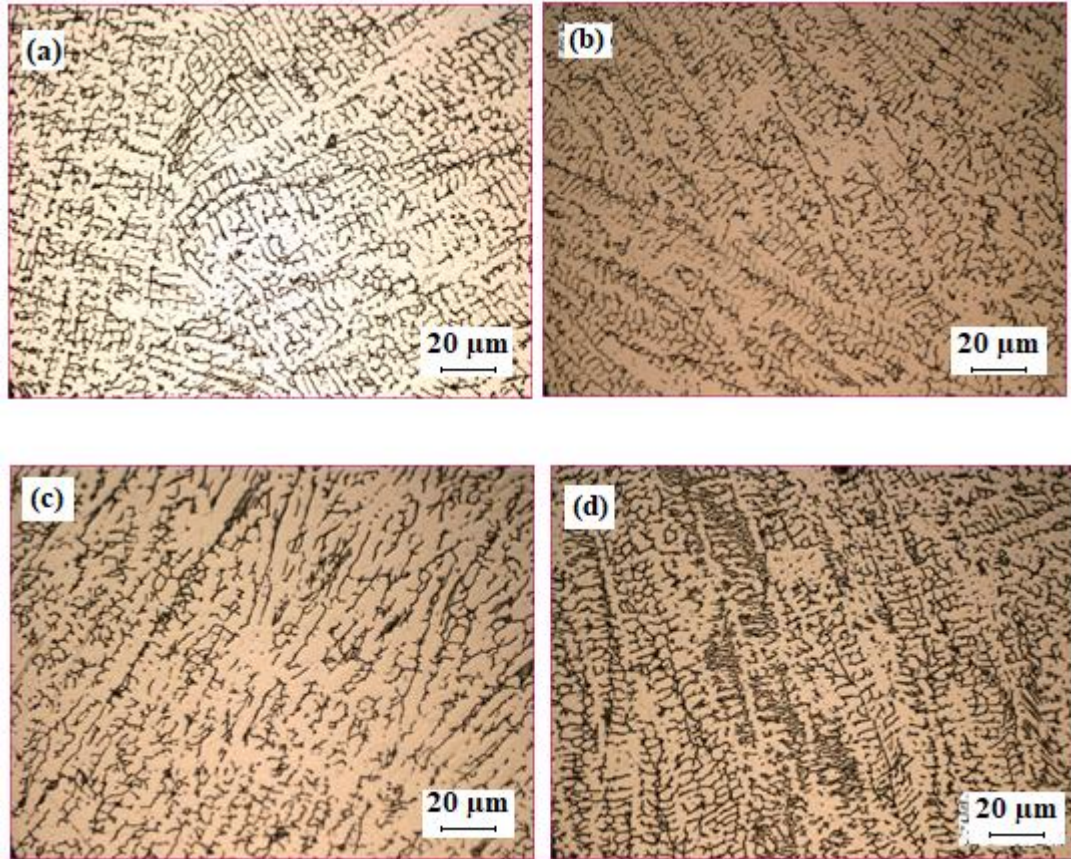


Figure 6.5 Weld metal microstructure of 316L ASS steel tested at (a) 200°C, (b) 400 °C, (c) 600 °C and (d) 800°C.

6.3. TENSILE TEST RESULTS AND EVALUATION

The aim of this work was to examine dynamic strain ageing behaviour of 316L austenitic stainless steel under as-received, as-welded and weld metal conditions. Therefore as-received samples, as-welded samples and weld metal were hot tensile tested in the temperature range of 25-800°C at a strain rate of $1 \times 10^{-3} \text{ s}^{-1}$. The obtained results were evaluated at the below sections.

6.3.1. Evaluation of Tensile Test Results for As-received Samples

Table 6.1 shows dynamic strain ageing results of as-received samples by giving yield strength (YS 0.2%), ultimate tensile strength (UTS) and elongation (%) for the testing temperature in the range of 25-800°C. As can be seen that as-received samples revealed small variation in YS and UTS values up to 600°C. YS

and UTS values change drastically beyond 600°C. Elongation (%) showed a continuous decrease as the testing temperature increased to 400°C. Further increase in testing temperature of 500-800°C has increased the elongation (%). This suggests that dynamic strain ageing takes place in 316L austenitic stainless steel up to 600°C under as-received conditions. Dynamic strain ageing generally occurs due to interactions between mobile dislocations and diffusing solute atoms in solid solutions [59].

Table 6.1 Mechanical properties of as-received samples tested at temperatures of 25-800°C.

Test Temperatures (°C)	Yield Strength (0.2%, MPa)	Tensile Strength (MPa)	Elongation (%)
25	388	650	66
100	265	539	45
200	215	487	41
300	206	470	36
400	202	473	32
500	161	449	36
600	160	401	43
700	142	282	70
800	125	178	112

Figure 6.6 shows stress and strain diagrams for as-received samples in the temperature range of 25-800°C. As can be seen the stress and strain diagrams showed significant changes in appearance when the testing temperature was increased for the strain rate of $1 \times 10^{-3} \text{ s}^{-1}$. Serrated yielding, which is characteristic of dynamic strain ageing, was clearly observed in as-received samples at testing temperatures of 500-600°C. The serrations disappeared from the curves at higher temperatures of 700-800°C. Such serrated behavior is manifested by a plateau in the variation of strength and a minima in the variation of elongation. Dynamic strain ageing refers to the interactions between diffusing solute atoms and mobile dislocations. Deformation rate and temperature, which affect the diffusing solute atoms and the velocity of mobile dislocation, play very important

role for dynamic strain ageing. At a high deformation rate and low temperature, the diffusion of solute atoms is slower than dislocation for the occurrence of strain ageing. However, at a low deformation rate and high temperature, solute atoms can move with the dislocations resulted of dynamic strain ageing disappear. Therefore, dynamic strain ageing can be observed in the temperature range of intermediate strain rates and temperatures [27, 60].

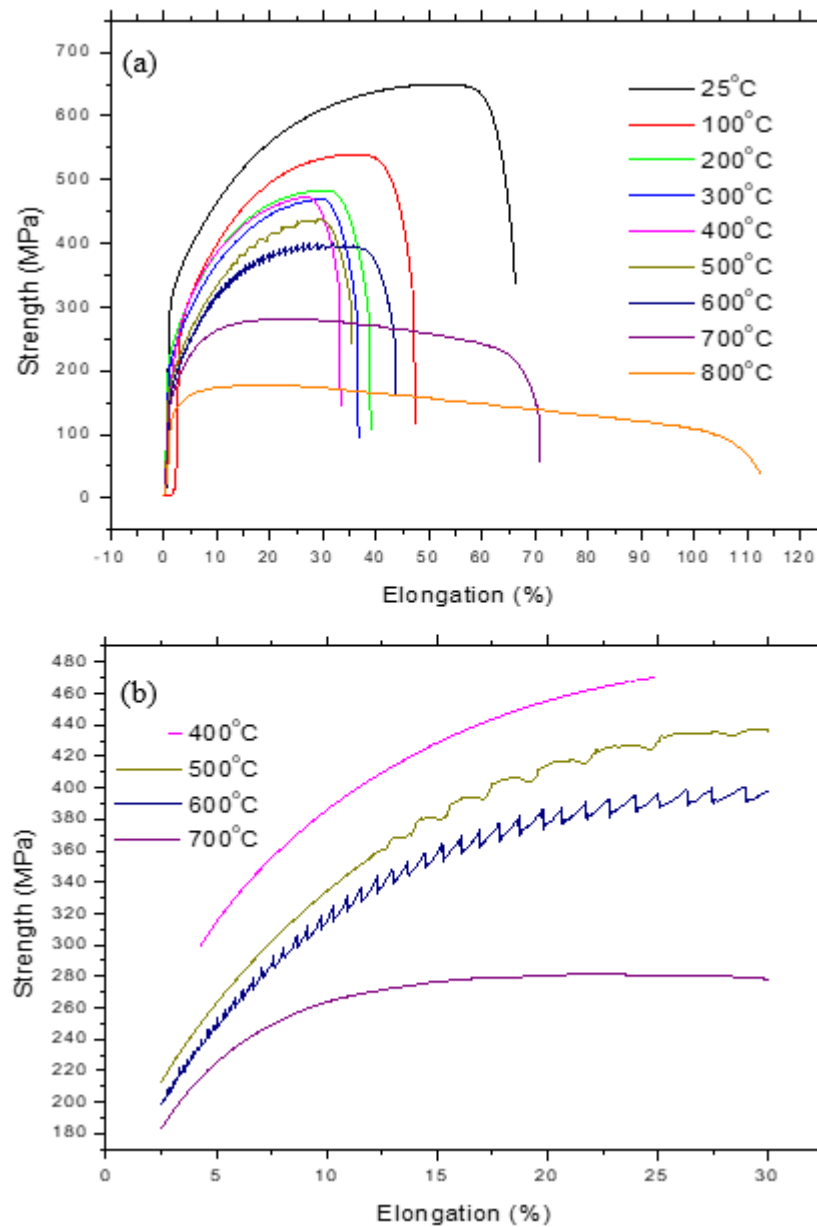


Figure 6.6 Stress and strain diagrams of as-received samples: (a) full range profiles and (b) sections in large scale.

6.3.2. Evaluation of Tensile Test Results for As-welded Samples

Table 6.2 shows dynamic strain ageing results of as-welded samples by giving yield strength (YS 0.2%), ultimate tensile strength (UTS) and elongation (%) for the testing temperature in the range of 25-800°C. As is seen as-welded samples showed slight decrease in YS and UTS values up to 600°C. YS and UTS values change drastically beyond 600°C. Elongation (%) showed a continuous decrease as the testing temperature increased to 600°C. Further increase in testing temperature of 700°C or 800°C has increased the elongation (%). This suggested that dynamic strain ageing occurred in 316L austenitic stainless steel up to 600°C under as-welded conditions. Dynamic strain ageing takes place in 316L as-welded samples due to interactions between mobile dislocations and diffusing solute atoms in solid solutions [59].

As a result of accumulation of solute atoms around the dislocation, the system would acquire the minimum strain energy. This leads to a situation where dislocation is anchored by the solute atom and would become immobile [61]. This interaction between dislocation and solute atom will be broken only at a higher temperature by thermal agitation [62]. It has been extensively reported that dynamic strain ageing plays an important role in the mechanical properties of 316 type stainless steels at temperatures of 200°C to 800°C [63]. The results obtained from present study is consistent with the results obtained from Prasanna and Udupa [64] who studied dynamic strain ageing in AISI 316L type stainless steel. It was shown that dynamic strain ageing is operative up to a temperature of 500°C.

Table 6.2 Mechanical properties of as-welded samples tested at temperatures of 25-800°C.

Test Temperatures (°C)	Yield strength (0.2%, MPa)	Tensile Strength (MPa)	Elongation (%)
25	368	635	34
100	306	544	36
200	254	487	30
300	269	481	31
400	283	473	28
500	260	444	28
600	231	409	27
700	181	286	52
800	138	179	80

The results also indicated that as-welded samples are more susceptible to dynamic strain ageing. For example, as-welded samples showed higher values in YS and UTS but lower values in elongation (%) compared to the as-received samples for all testing temperatures of 25-800°C. This is due to the least and greatest amount of free or uncombined interstitial atoms such as carbon and nitrogen. In Table 6.2, it is seen that the dynamic strain ageing depends on the presence of free interstitial atoms in solid solution. Incomplete precipitation of carbides and nitrides occurs during fast cooling after welding, which resulted in an increase in the amount of uncombined carbon and nitrogen in solid solution in the as-welded condition. Higher dislocation density in the HAZ of the welds can also be seen owing to development of strains during the thermal cycle of welding and in later loading in the dynamic strain ageing temperature range, e.g., inside of the crack tip plastic zone [65, 66].

Dolby and Saunders [67] showed that steels with low interstitial nitrogen content and a fine ferrite grain size exhibit a low susceptibility to embrittlement due to dynamic strain ageing in sub-critical HAZ regions of weldments. Dynamic strain ageing has also been observed in the root regions of high strength low alloy steels welds by Blake [68] due to locking of dislocations by carbon and nitrogen atoms under the weld root deformation and cooling stresses. In fact Cochrane et al. [69] have shown

dynamic strain ageing to be the dominant factor in weld root toughness deterioration. While the problem of strain ageing embrittlement is adequately investigated long since from the steel makers points of view, there appears to be very limited data on this problem with respect to steels and weld metals relevant to offshore fabrications [70].

Figure 6.7 shows stress and strain diagrams of as-welded samples in the temperature range of 25-800°C. As can be seen, the stress and strain diagrams showed significant changes in appearance when the testing temperature was increased for the strain rate of $1 \times 10^{-3} \text{ s}^{-1}$. Serrated yielding which is the characteristic of dynamic strain ageing was clearly seen in as-welded samples at testing temperatures of 500°C and 600°C. The serrations disappeared from the curves at higher temperatures of 700°C and 800°C. Such serrated behaviour is shown by a plateau in the variation of strength and a minima in the variation of elongation. Deformation rate and temperature, which effect the diffusing solute atoms and the velocity of mobile dislocation, play very important role for dynamic strain ageing.

At high deformation rate and low temperature, the diffusion of solute atoms is more slower than dislocation for the occurrence of strain ageing. However, at low deformation rate and high temperature, solute atoms can move with the dislocations resulted of dynamic strain ageing disappear. Dynamic strain ageing occurred during the interaction of mobile dislocation and solute atoms and waiting time in which dislocations are arrested by local obstacles such as precipitate particles in the glade plane [60, 71]. It is clear from foregoing discussion that dynamic strain ageing takes place in 316L austenitic stainless steel under as-welded condition results from the interactions of solute atoms and dislocation in the temperature range of 200-600°C.

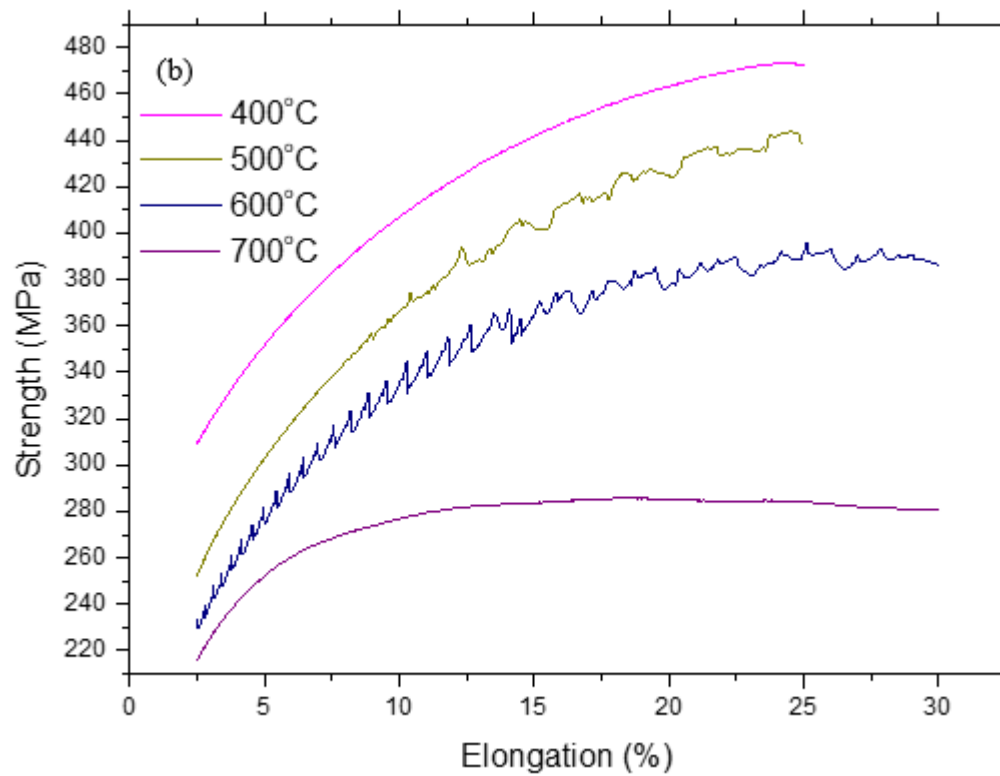
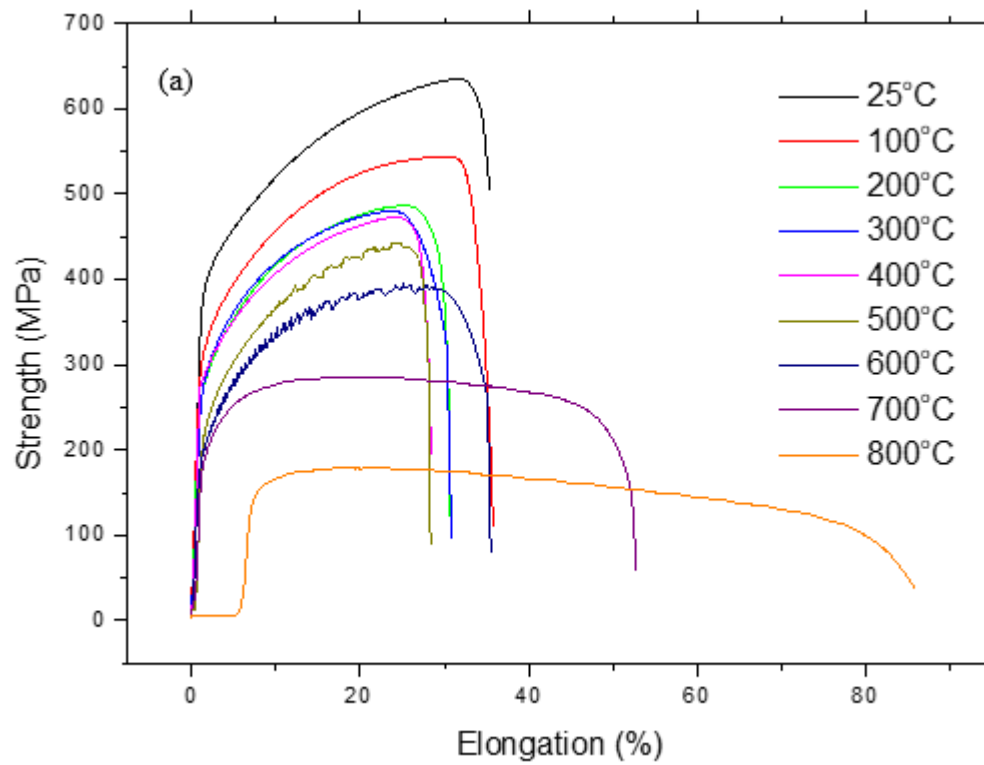


Figure 6.7 Stress and strain diagrams of as-welded samples: (a) full range profiles and (b) sections in large scale.

6.3.3. Evaluation of Tensile Test Results for Weld Metal Samples

Dynamic strain ageing results of weld metal are seen in Table 6.3 which gives YS (0.2%), UTS and elongation (%) of samples tested at 25-800°C. As can be seen that weld metal showed small differences in YS and UTS up to 600°C, but it revealed drastic change when the test temperature was increased beyond 600°C. A decrease in elongation (%) values of weld metal was observed up to 600°C. A further increase in test temperature of 700°C or 800°C increased the elongation (%). This indicates that weld metal of 316L austenitic stainless steel is susceptible to dynamic strain ageing up to 600°C. This is due to the the presence of free carbon and nitrogen atoms in solution. The diffusion of solute atoms to dislocations will prevent the movement of the dislocations resulted in higher strength and lower ductility in steel [59, 65].

Heating and cooling rates in fusion welding are usually high and the heated metal subject to plastic strain during cooling. For this reason, fast cooling after welding does not allow full precipitation of carbides and nitrides. This resulted in an increase in the amount of uncombined carbon and nitrogen in solution in the weld metal which can cause dynamic strain ageing to occur. Dynamic strain ageing is considered to contribute to higher strength in weld metal by effectively increasing the dislocation density for a given strain. A high dislocation density, the fact that many dislocations will be pinned by solute atoms and that bowed dislocations need higher stresses for unpinning, mean that strength in the weld metal will be higher [72].

Table 6.3 Mechanical properties of weld metal tested at temperatures of 25-800°C.

Test Temperatures (°C)	Yield Strength (0.2%, MPa)	Tensile Strength (MPa)	Elongation (%)
25	395	659	57
100	349	559	37
200	407	543	31
300	389	505	20
400	396	521	24
500	336	485	25
600	307	423	22
700	190	295	37
800	153	184	81

Stress and strain diagrams of weld metal tensile tested at temperatures of 25-800°C are seen in Figure 6.8. As is seen, stress and strain diagrams revealed serrated behaviour when the testing temperatures increased at a strain rate of $1 \times 10^{-3} \text{ s}^{-1}$. Serrated behavior which is the manifestation of dynamic strain ageing was observed during tensile testing at 500°C and 600°C. The serrations disappeared from the curves when the samples were tested at 700°C or 800°C. Such serrated behaviour occurs when the dislocations and solute atoms interact which may become an important controlling factor for strengthening. The interaction between dislocations and solute atoms causes the appearance of serrations on the curve, an increase in tensile strength and strain hardening, but a decrease in ductility. Dynamic strain ageing will occur when the interstitial atoms diffuse and pin the dislocations. As a result of this, serrations occur because of rapid generation of new dislocations which are needed to sustain flow [73].

Deformation rate and temperature, which are very important for dynamic strain ageing, effect the diffusing solute atoms and the velocity of mobile dislocations. The presence or absence of serrations is governed by strain rate at a given temperature. For example, dislocation diffusion is higher than solute atoms for the occurrence of dynamic strain ageing at low temperature and high deformation rate. However, dynamic strain ageing disappears at high temperature and low deformation rate,

because dislocations can move with the solute atoms. Therefore, dynamic strain ageing can occur at the temperatures of intermediate strain rates [60]. It is clear from foregoing discussion that dynamic strain ageing occurs in weld metal of 316L austenitic stainless steel due to the interactions between dislocations and solute atoms in the temperature range of 200-600°C.

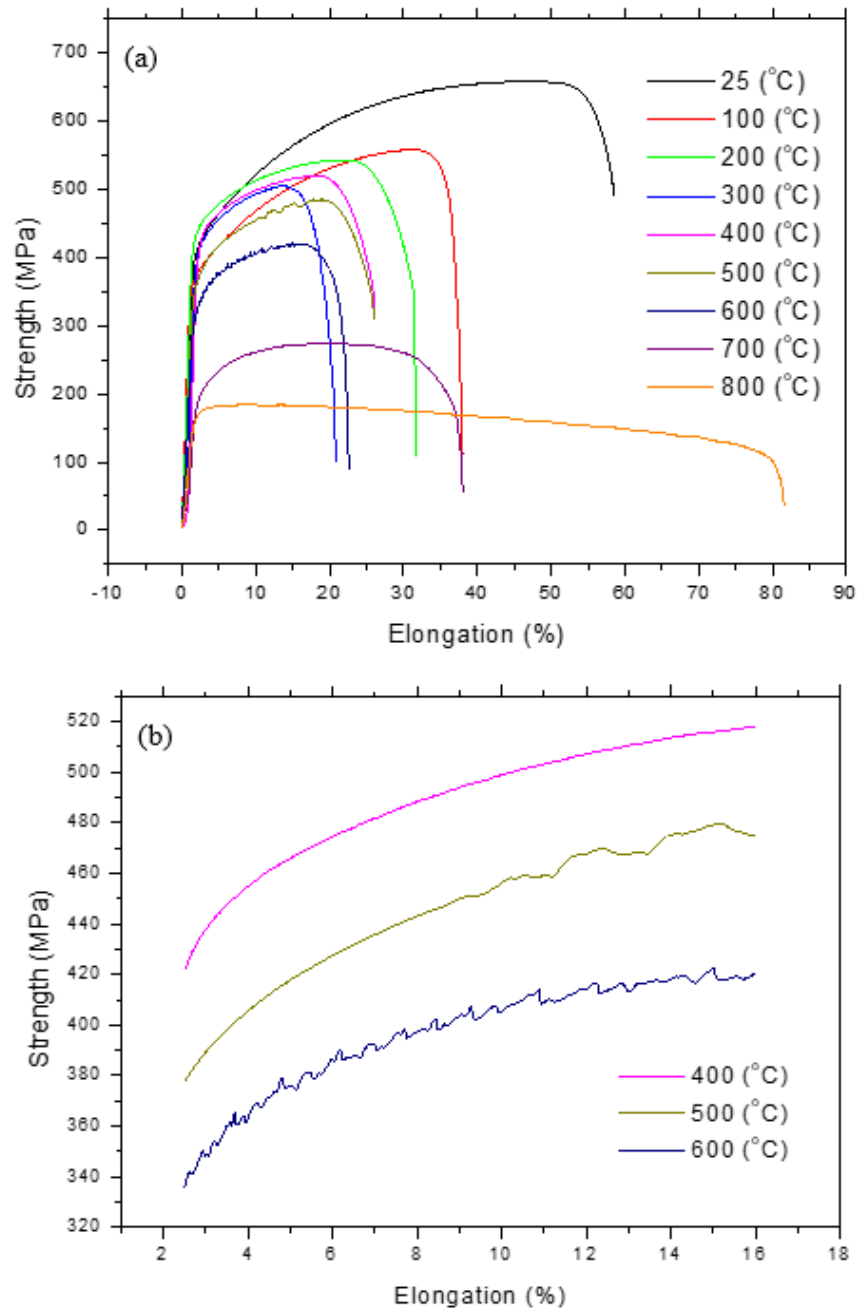


Figure 6.8 Stress and strain diagrams of weld metal samples: (a) full range profiles and (b) section large scale.

6.3.4. Evaluation of Hardness Test Results for As-received Samples, As-welded Samples and Weld Metal Samples

Figure 6.9 shows Vickers microhardness measurement of the base metal, HAZ and weld metal. It was observed that the highest hardness was obtained in the weld metal. This could be the presence of higher amount of Cr in solid solution due to welding thermal cycle. It was shown that higher Cr content in the weld metal increases hardenability of steel [74, 75]. This is consistent with the results obtained by Ohkubo et al. [76] and Hee-jin and Hae-woo [77] who investigated the role of alloying elements on the mechanical performance of austenitic stainless steel. Higher hardness in weld metals may also be owing to melting and solidification of the area during or cooling after welding. Generally, as-welded samples revealed the lowest and highest hardness in the base metal and weld metal respectively. Hardness of HAZ was found to be higher than base metal, but lower than weld metal.

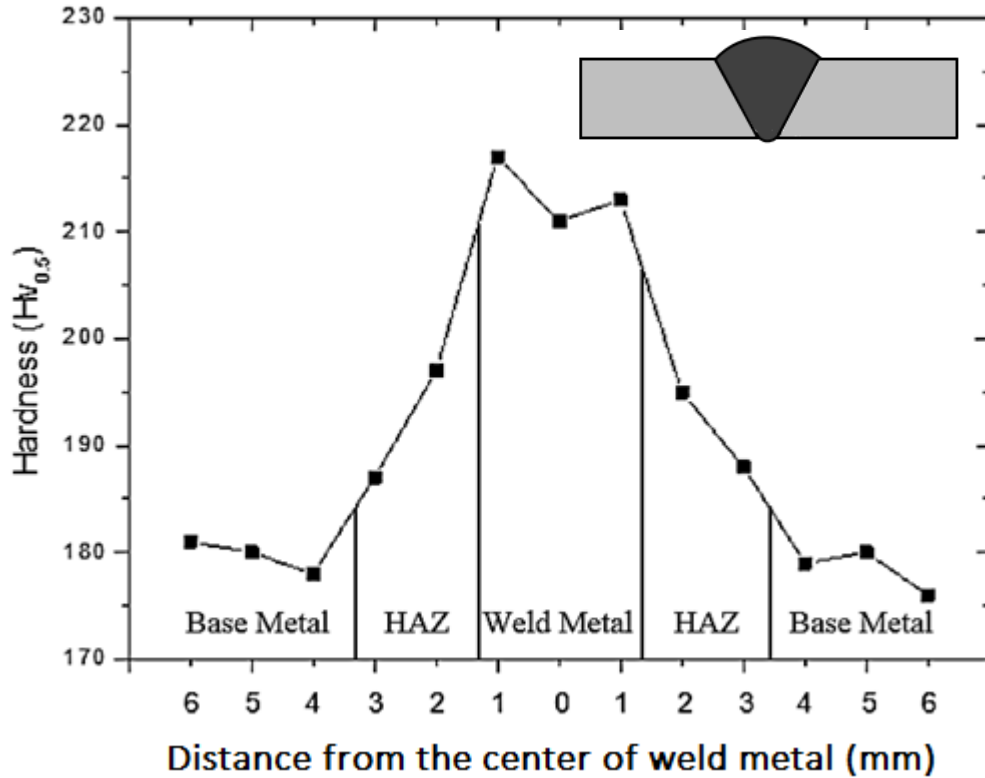


Figure 6.9 Vickers hardness (HV0.5) in the welded sections after welding.

6.3.5. Comparative Evaluation of Tensile Test Results for As-received Samples, As-welded Samples and Weld Metal Samples.

Workhardening index (n) which is used to determine the occurrence of dynamic strain ageing was evaluated according to the Holloman equation for as-received samples, as-welded samples and weld metal. Detail of Holloman equation was given in section 5.5. The evaluated n values following equation 1 as a function of hot tensile test temperatures are seen in Figure 6.10. As can be seen, workhardening index increased in the temperature range of 400-600°C for as-received samples, as-welded samples and weld metal. This indicates the interaction between dislocation and carbon or nitrogen atoms which affects the work hardening behaviour of 316L austenitic stainless steel under as-received, as-welded and weld metal conditions.

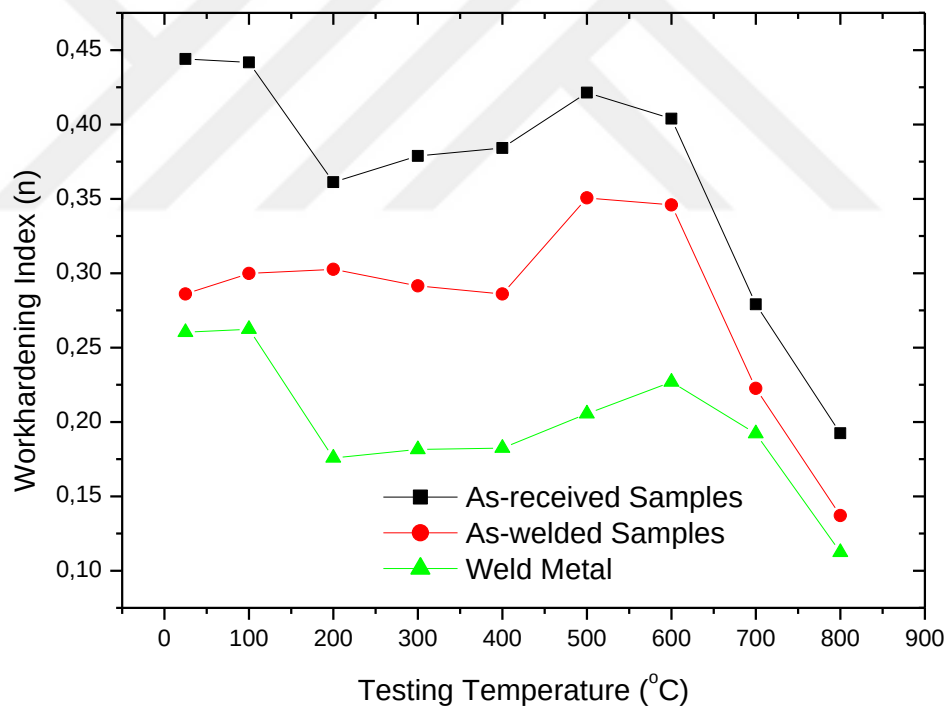


Figure 6.10 Workhardening index (n) values of the as-received samples, as welded samples and weld metal tested at different temperatures.

The increased workhardening index are believed to arise from greater than normal dislocation densities in steels that exhibit dynamic strain ageing. These high dislocation densities occur due to the pinning, which requires fresh dislocations to maintain the applied strain rate. Tensile testing at 400-600°C results in a stronger

interaction between solute atoms and dislocation which increased the workhardening of a material, since the deformation at 400-600°C can produce greater number of dislocations due to the interaction of dislocations and interstitial solute atoms (carbon and nitrogen) contributed to stronger workhardening [60]. Kaçar and Gündüz [78] showed that the activation energy for strain ageing in AISI 430 austenitic stainless steel is bigger than plain carbon steel.

The activation energy for strain ageing in AISI 430 stainless steel was also calculated as $\Delta H = 126.7 \text{ kJmol}^{-1}$ by Buona et al. [79]. This value is bigger than the activation energy for strain ageing in low carbon steels which is equal to the activation energy for the diffusion of C atoms in ferrite, 84.2 kJmol^{-1} . However, it can be said that the affinity of chromium for carbon atoms changes the activation energy for diffusion of the latter in 316L austenitic stainless steel to such an extent. Therefore, as-received samples, as-welded samples and weld metal did not show any discontinuous yielding behaviour, because chromium retards the diffusion of carbon to dislocations at lower temperatures in 316L austenitic stainless steel.

Figure 6.10 also shows that the as-received samples achieved higher workhardening index n values than as-welded samples and weld metal due to good formability. The as-received samples reached a larger reduction in workhardening index (5%) while as-welded samples showed an increase in workhardening index (23%) at the peak ageing temperature of 500°C, compared to those in the room temperature testing conditions. This indicates that dynamic strain ageing is more pronounced in as-welded samples compared to the as-received samples. It was also found that weld metal showed the lowest workhardening index for all testing temperatures compared to the as-received samples and as-welded samples due to poor formability which is resulted in dynamic strain ageing.

This is also confirmed in Figure 6.11, where YS is graphed against testing temperature. As seen in Figure 6.11, there is a decrease in YS values of as-welded samples and weld metal with increasing temperature, reaching a minimum at 100°C or 200°C; it then increased with increasing temperature and reached a maximum at 400°C. This indicated the interaction between carbon and/or nitrogen with

dislocation which prevented dislocation movement. YS values changed drastically above 600°C. This also shows that dynamic strain ageing is operative in as-welded samples and weld metal at the test temperatures of 25-800°C. It is proposed that dynamic strain ageing occurs when the dislocations are temporarily held up at local obstacles in the glide plane [80]. However, as-received samples revealed a decrease in YS value up to 400°C, after which it decreased too fast. This also suggested that as-welded samples and weld metal are more affected by dynamic strain ageing compared to the as-received samples at the test temperatures of 25-800 °C.

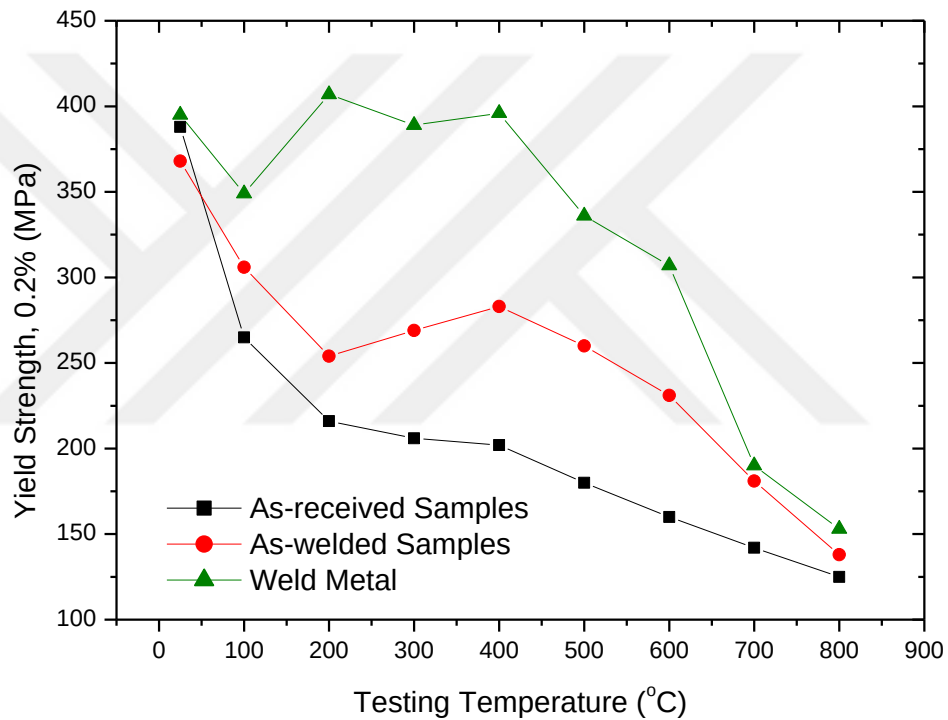


Figure 6.11 Yield strength values of as-received samples, as-welded samples and weld metal tested at different temperatures.

The effect of testing temperature on UTS is also presented in Figure 6.12 for as-received samples, as-welded samples and weld metal. As can be seen, UTS continuously decreased with increasing in testing temperature for as-received samples and as-welded samples from 25°C to 200°C at the strain rate of $1 \times 10^{-3} \text{s}^{-1}$. There was flattening of the plot in the temperature of 200°C to 600°C at the strain rate of $1 \times 10^{-3} \text{s}^{-1}$. However, weld metal showed an increase in UTS at the temperature of 400°C. Further increase in testing temperatures continuously decreased UTS. Taştēmür and Gündüz [81] investigated tensile properties of AISI H10 hot work tool

steel in the temperature range of 25-700°C under as-received and as-quenched conditions. They observed that the UTS of the as-received samples decreased with increasing temperature from 25-400°C and then rapid decrease was observed with further increase in temperature. However, as-quenched samples showed an increase in UTS at the temperature of 25-200°C. Further increase in testing temperatures continuously decreased UTS.

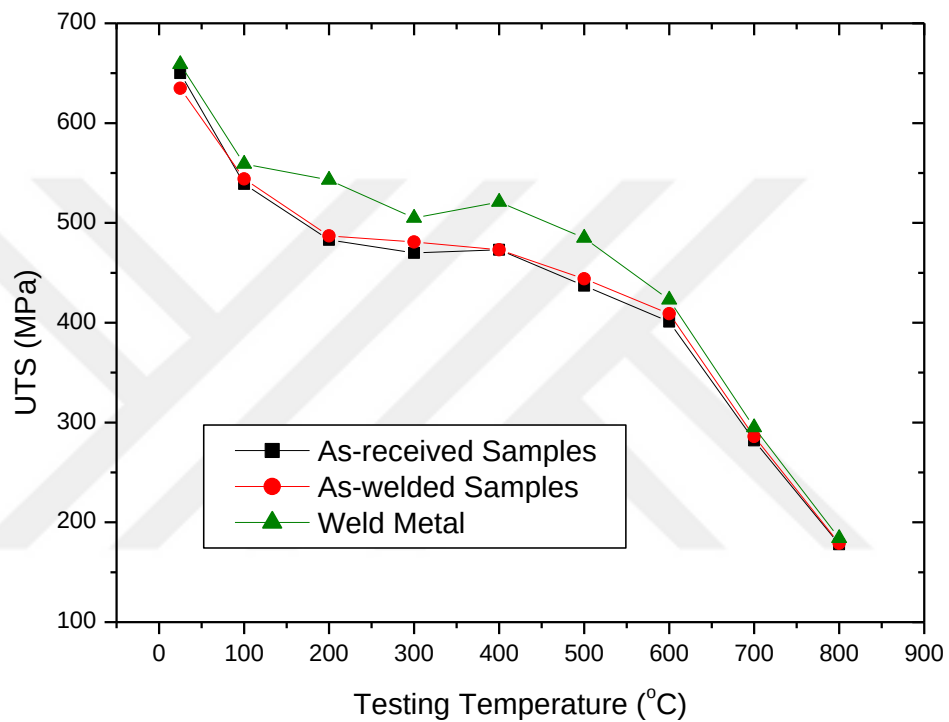


Figure 6.12 Ultimate tensile strength values of as-received samples, as-welded samples and weld metal tested at different temperatures.

The variation of percentage elongation with temperature is shown in Figure 6.13. As is seen, elongation of as-received samples, as-welded samples and weld metal showed a decrease in the dynamic strain ageing temperature range (200-600°C). Correspondingly, the ductility decreases with increasing temperature, reaching minimum values at around 200-400°C, and then increases with increase in test temperatures. This means that the effect of dynamic strain ageing is dominant at this temperature range. Also as-welded samples and weld metal showed lower percentage elongation than as-received samples which indicated that dynamic strain ageing is more operative in as-welded samples and weld metal compared to the as-received samples. Gupta et al. [82] have shown in their study on elevated temperature tensile

properties of a 3Cr-1Mo steel that the ductility decreases with increasing temperature reaching minimum values at around 350-400°C, and then increases with increase in test temperatures.

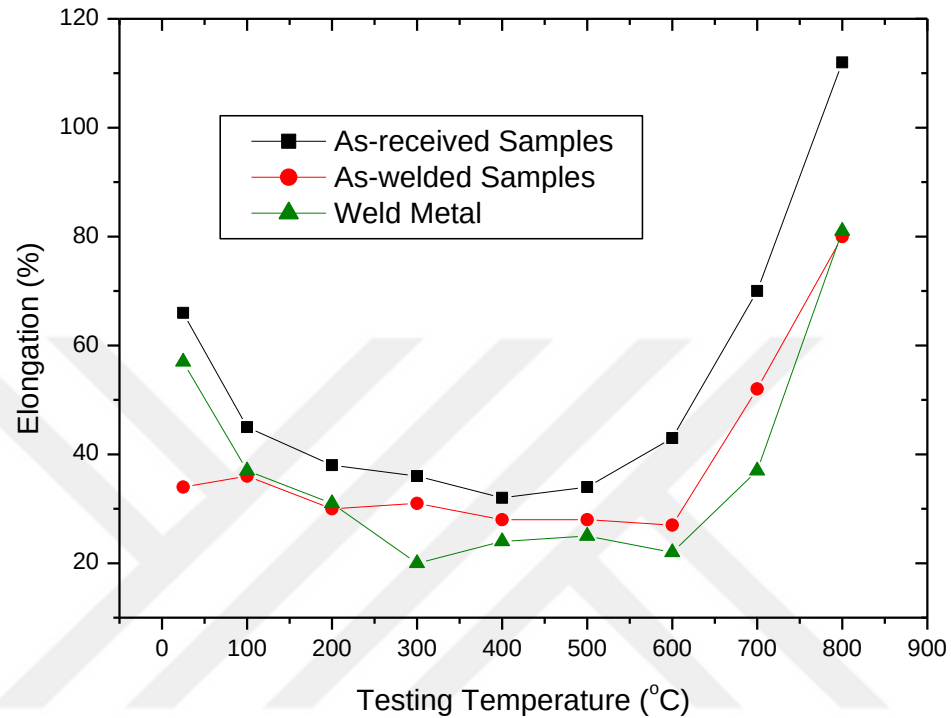


Figure 6.13 Elongation to fracture values of as-received samples, as-welded samples and weld metal tested at different temperatures.

Tensile test results obtained from present study indicated that as-welded samples revealed lower values in elongation (%) but higher values in YS and UTS than the as-received samples at the temperatures of 25–800°C. This indicated that that as-welded samples are more effected by dynamic strain ageing than the as-received samples. However, the tensile test results of weld metal showed that weld metal is more susceptible to dynamic strain ageing than the as-received and as-welded samples with evidence more increase in strength but decrease in elongation (%) at the testing temperatures of 25-800°C. This is due to the presence of higher Cr content in weld metal compared to that of as-received samples, because weld wire Cr content is about 18.5 wt. % which is higher than as-received samples. It is generally accepted that the elevated Cr content of the weld metal increases the strength of the steel [54, 77]. Higher strength of weld metal could also be due to heat input during welding,

melting and solidification of the area as suggested by Lippold and Kotecki [75] who investigated welding metallurgy and weldability of stainless steel.

As discussed in the sections 6.3.1-6.3.3, as received samples, as-welded samples and weld metal showed serrations on the stress and strain curves after tensile testing at 500°C or 600°C. Serrated yielding refers to the occurrence of repeated discontinuous yielding during plastic deformation. The discontinuous yielding is evidenced on a stress-strain curve by the appearance of serrations usually after reaching a critical level of plastic strain. On the basis of the appearance of serrations on the stress-strain diagrams, five types of serrated yielding (A, B, C, D and E types) are often observed which is discussed in section 3.3.2 (see Figure 3.3 in section 3.3.2) [26].

Type A, B, D, E serrations are observed in the stress-strain diagrams of 316L stainless steel under as-received conditions. Type A and type D serrations occurred at low and high strain rates respectively for the testing temperature of 500°C. In test conducted at 600°C, the amplitude of the serrations raised with strain and they changed from types A+D to types B+E. Type B serrations occurred at low strain and change over to type E serration at high strain. The tensile test results indicated that serrated behaviour was observed in 316L austenitic stainless steel at testing temperatures of 500 and 600°C. However serrated behaviour was not observed above 600°C. Different solute atoms and mechanisms can participate in the interactions of solute atom and mobile dislocation at different temperatures [83, 84].

The dynamic strain ageing, observed at temperature of 500°C, shows itself as serrated flow of types A and D. Interaction of interstitial atoms such as carbon and nitrogen with dislocations during deformation is considered to be responsible for the dynamic strain ageing in austenitic stainless steels. At temperatures about 600°C, where the type B and E serrations are seen, the dynamic strain ageing can be attributed to interaction of substitutional solute atoms such as Cr or Mo with mobile dislocations [85]. Mannan et al. [86], also observed type A, B, and C serrations in 316 stainless steel.

The temperature range, where the serrations is observed, is similar with as-received samples for as-welded samples, strained at the same strain rate of $1 \times 10^{-3} \text{ s}^{-1}$. Types A and D serrations were observed on stress-strain curves obtained at strain rate of $1 \times 10^{-3} \text{ s}^{-1}$ at 500°C for as-welded samples. At 600°C serrated flow changes to type C for as welded samples.

Type D serrations are observed in the stress–strain diagrams of 316L austenitic stainless steel weld metal at high strain rate for the testing temperature of 500°C. In test conducted at 600°C, the amplitude of the serrations raised with strain and they changed from types D to types A+E. Type A serrations occurred at low strain and change over to type E serration at high strain. The tensile test results indicated that serrated behaviour was observed in 316L austenitic stainless steel at testing temperatures of 500°C and 600°C. However serrated behaviour was not observed above 600°C. Testing at different temperatures effects the interactions of solute atom and mobile dislocation. As a result of this steels show different type serrations depending on temperature and strain rates [85].

When the serrations behaviour results of weld metal are compared with the serrations behaviour results of the as-received and as-welded samples of 316L austenitic stainless steel, it was observed that weld metal revealed more pronounced serrations than as-received and as-welded samples owing to the presence of higher amount of solute atoms in solution [87]. Dissolution of carbides occurs in the weld metal results in larger amount of solute atoms in the solution. This cause more pronounced serrations in weld metal than as-received and as-welded samples. As-welded samples also showed more pronounced serrations behaviour than as-received samples. Partial dissolution of carbides occurs in the HAZ next to the fusion line during welding. This results in a larger amount of solute atoms in the solid solution which can result in more pronounced serrations in as-welded samples compared to the as-received samples.

6.4. SCANNING ELECTRON MICROSCOPE (SEM) RESULTS AND EVALUATION

In this part, detailed microstructure and fracture surface analyses were carried out for the as-received samples, as-welded samples and weld metal which were hot tensile tested at temperatures of 25°C, 200°C, 400°C, 600°C and 800°C for a strain rate of $1 \times 10^{-3} \text{ s}^{-1}$. Detailed microstructure and fracture surface analyses were done by using SEM.

6.4.1. Microstructure Analyses of As-received Samples, As-welded Samples and Weld Metal Samples

6.4.1.1. Microstructure Analyses of As-Received Samples

Figures 6.14-6.18 show the scanning electron micrographs of the as-received samples tensile tested at different temperatures. Figures 6.14, 6.15, 6.16, 6.17 and 6.18 show the microstructures near fracture surface of the as-received samples tested at 25°C, 200°C, 400°C, 600°C and 800°C respectively. Martensite structure was observed in samples tested 25°C as seen in Figure 6.14. Deformation martensite occurs at the deformation twins, grain boundaries and shear bands which have large strain gradients. It was indicated that mechanical twins are the primary nucleation sites for deformation martensite [88].

The occurrence of deformation martensite is related to the austenite instability at temperatures close or below room temperature. The structural transformation susceptibility is correlated to the stacking fault energy (SFE), which is a function not only of the chemical composition, but also of the testing temperature. Austenitic stainless steels have high plasticity and can be easily cold formed. However, the hardening phenomena always occurs during cold processing. Nevertheless, the deformation martensite transformation can increase the workhardening rate and it may or may not be in favour of further material processing [89]

The results obtained from the present study suggest that the martensite develops at the shear bands. Ren-bo et al. [90] indicated that 316L austenitic stainless steel has low stacking fault energy (64 mJ/m^2), and a low stacking fault energy is generally related to higher susceptibility to martensite transformation. Many parallel shear bands can be seen in Figure 6.15, 6.16, 6.17 and 6.18 for the samples tested at 200°C, 400°C, 600°C and 800°C.

Figures 6.14-6.18 also reveals EDS analysis with the different spectrum points marked on the microstructure of samples tested at 25-800°C. For example, Figure 6.17 shows EDS analysis with the spectrum points 1, 2, 3 and 4 marked on the microstructure of samples tested at 600°C. Points 1 and 2 contain Fe, C, Cr and Mo, but point 3 contains Fe, C, Cr, O and Mo. The presence of these elements indicated that (Cr,Mo)C and CrO occurred during testing or cooling after testing at 600°C in which the sample showed serration behavior.

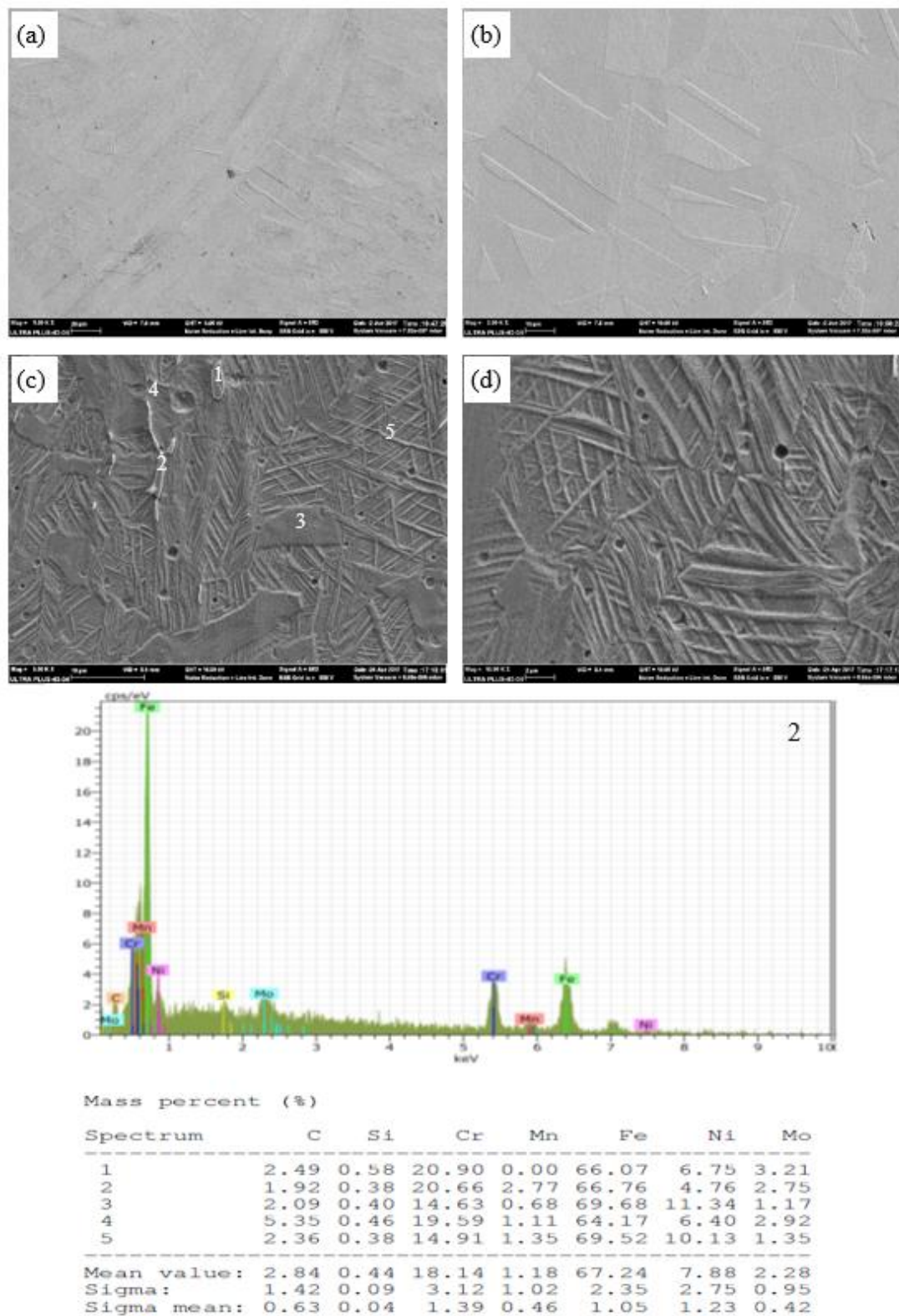


Figure 6.14 SEM micrographs with different magnification near fracture surface of as-received samples tested at 25°C with EDS analyses.

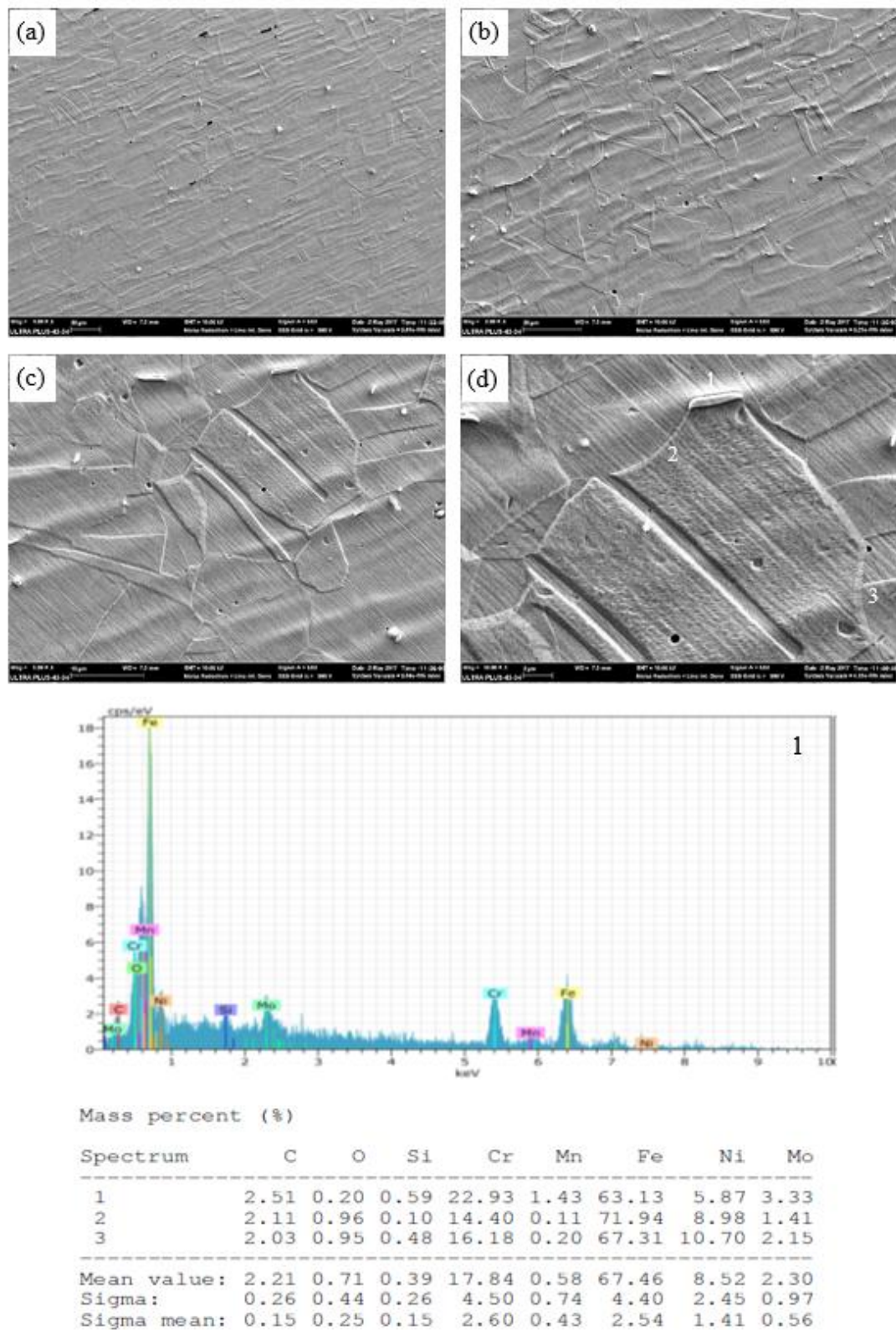


Figure 6.15 SEM micrographs with different magnification near fracture surface of as-received samples tested at 200°C with EDS analyses.

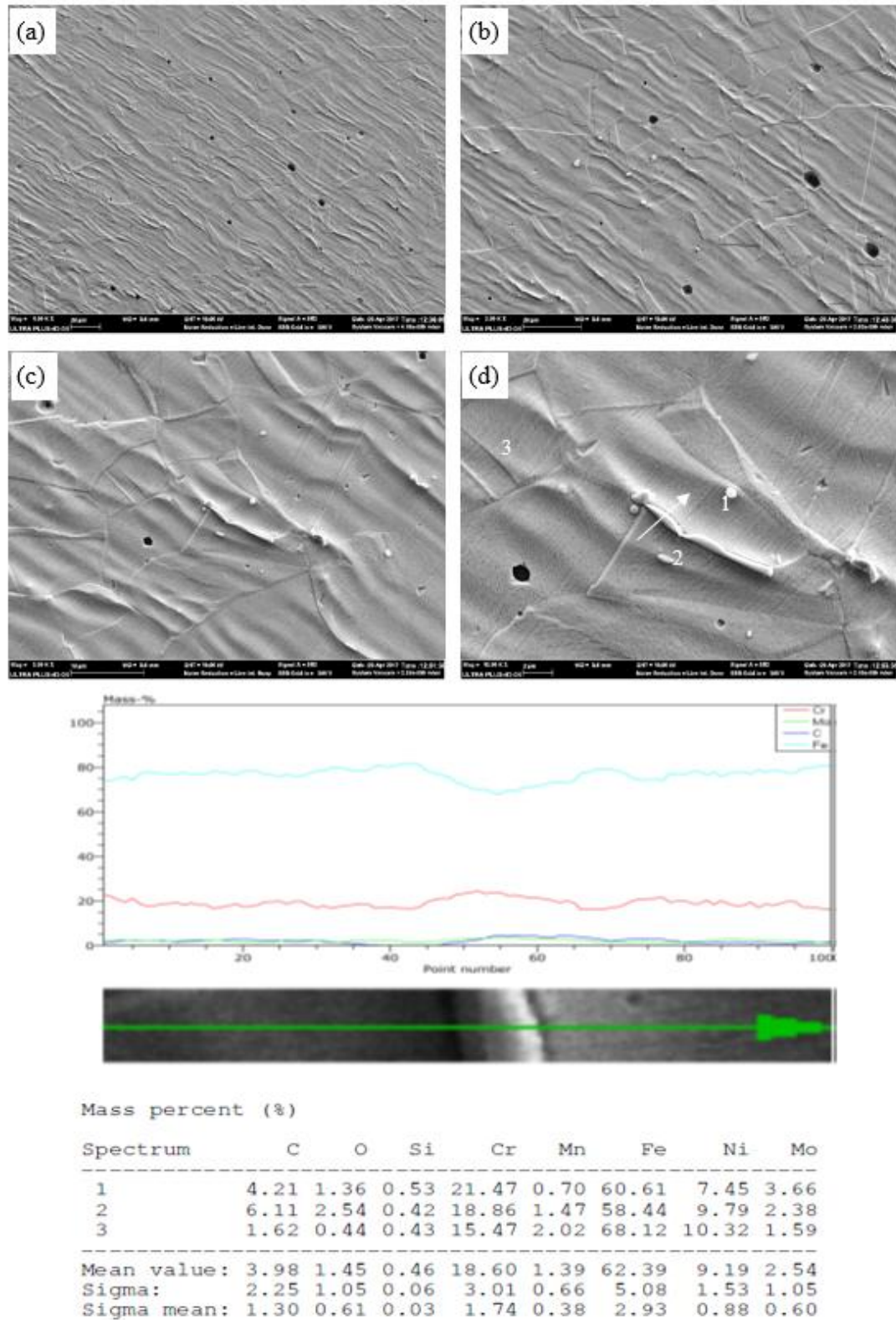


Figure 6.16 SEM micrographs with different magnification near fracture surface of as-received samples tested at 400°C with EDS analyses.

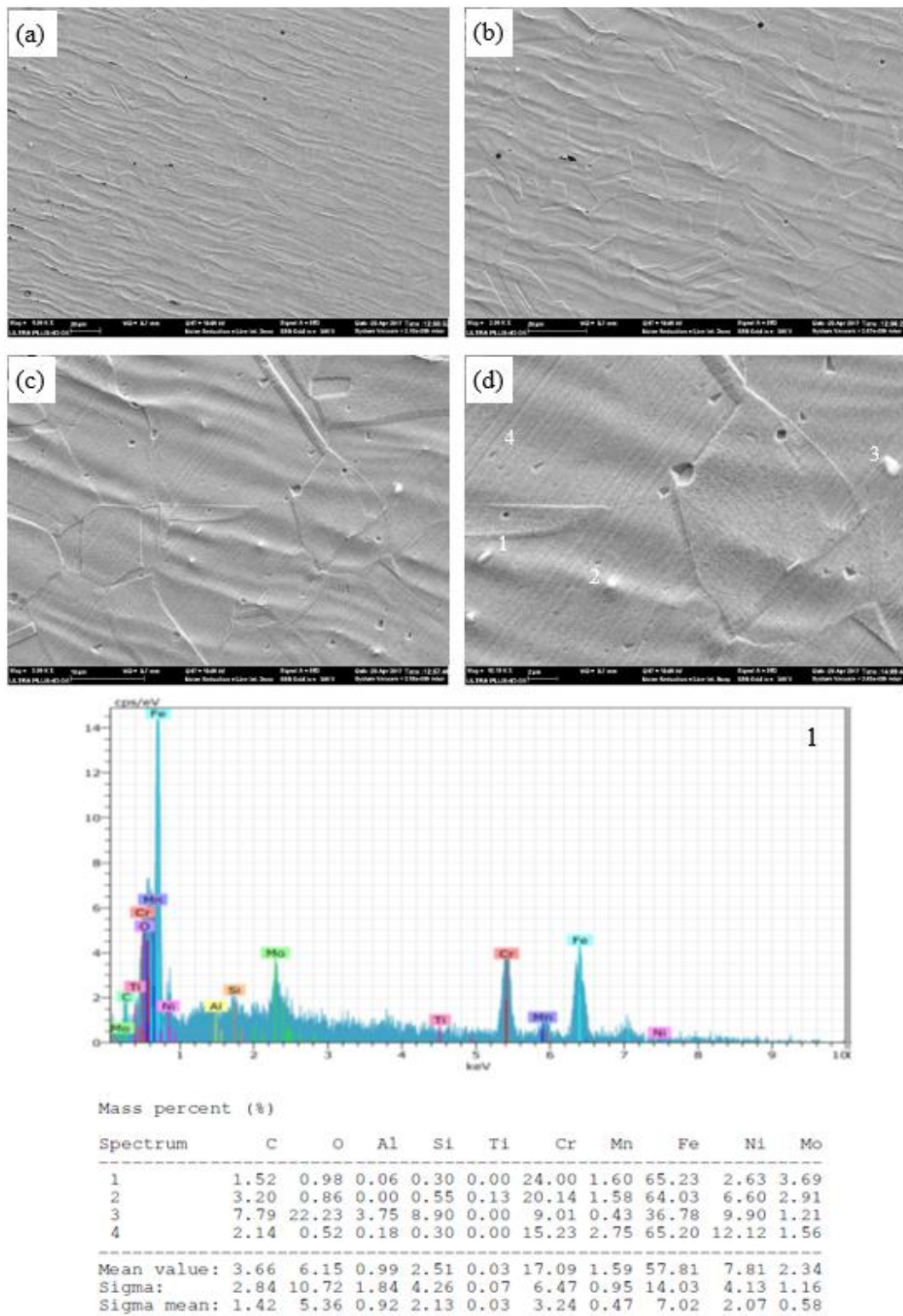


Figure 6.17 SEM micrographs with different magnification near fracture surface of as-received samples tested at 600°C with EDS analyses.

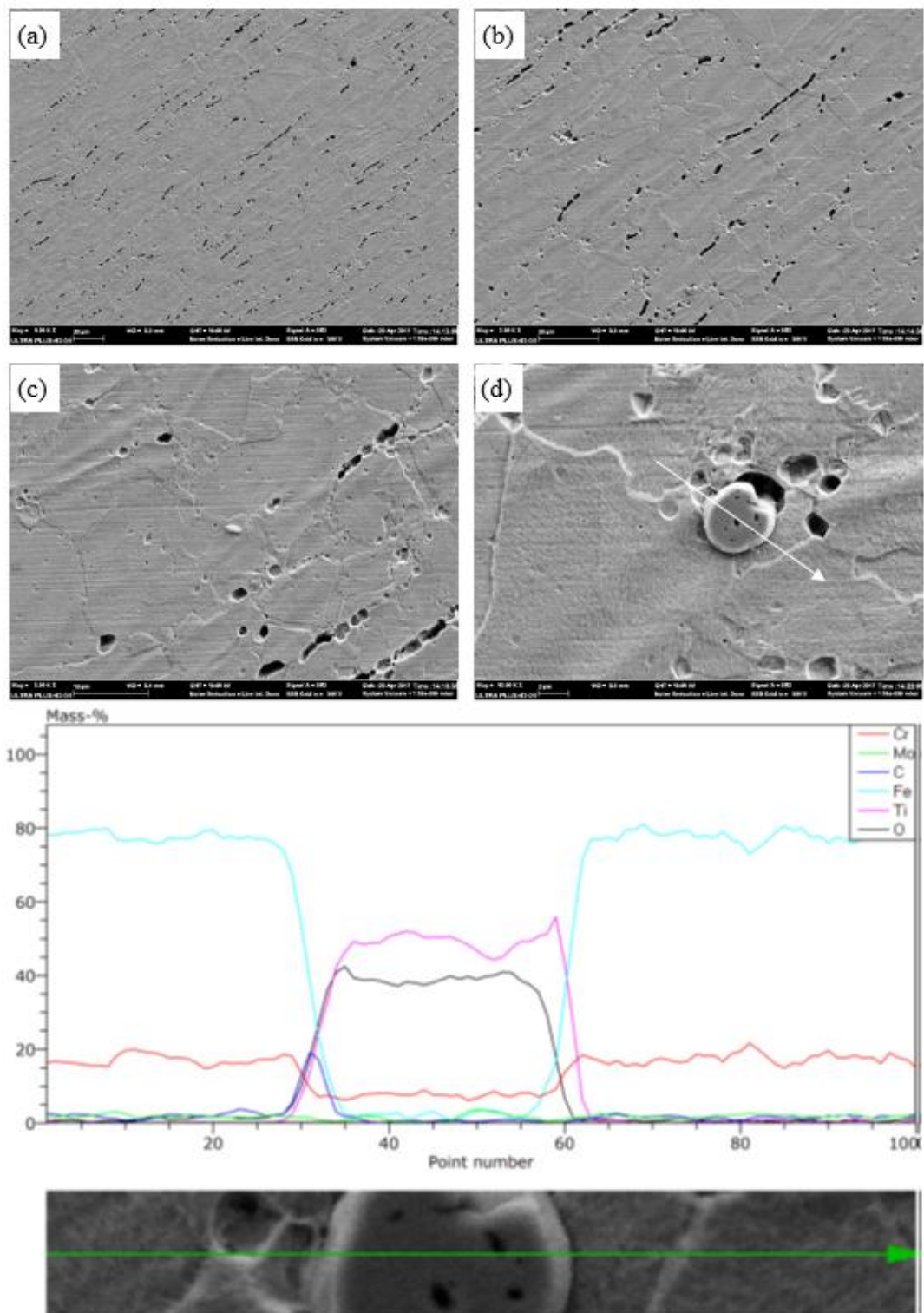


Figure 6.18 SEM micrographs with different magnification near fracture surface of as-received samples tested at 800°C with line EDS analyses.

6.4.1.2. Microstructure Analyses of As-Welded Samples

Figures 6.19-6.23 show the scanning electron micrographs of the as-welded samples tensile tested at different temperatures. Figures 6.19, 6.20, 6.21, 6.22 and 6.23 show the microstructures near fracture surface of the as-welded samples tested at 25°C, 200°C, 400°C, 600°C and 800°C respectively. Martensite structure was observed on shear bands of as-welded samples tested at 25°C as seen in Figure 6.19. Samples tested at 400°C, 600°C and 800°C also showed many parallel shear bands as seen in Figures 6.21, 6.22 and 6.23. This indicates that the deformation martensite develops at the shear bands which have large strain gradients [88].

Figure 6.20 revealed typical weld metal microstructure which consisted of skeletal ferrite in austenitic matrix with some carbides. The fracture of the as-welded samples tested at temperatures of 25-800°C started from adjacent to the fusion zone in the HAZ and continued through into the weld metal. The precipitation of carbides such as $M_{23}C_6$, MC, M_6C and M_7C_3 ($M = Cr, Mo$ and Fe) occurs in austenitic stainless steel during welding [55].

Figures 6.19-6.23 also reveals EDS analysis with the different spectrum points marked on the microstructure of samples tested at 25-800°C. For example, Figure 6.20 shows EDS analysis with the spectrum points 1, 2 and 3 marked on the microstructure of samples tested at 200°C. Points 1, 2 and 3 contain Fe, C, Cr and Mo. The presence of these elements indicated that CrC, MoC or (Cr, Mo)C occurred during testing or cooling after testing at 200°C.

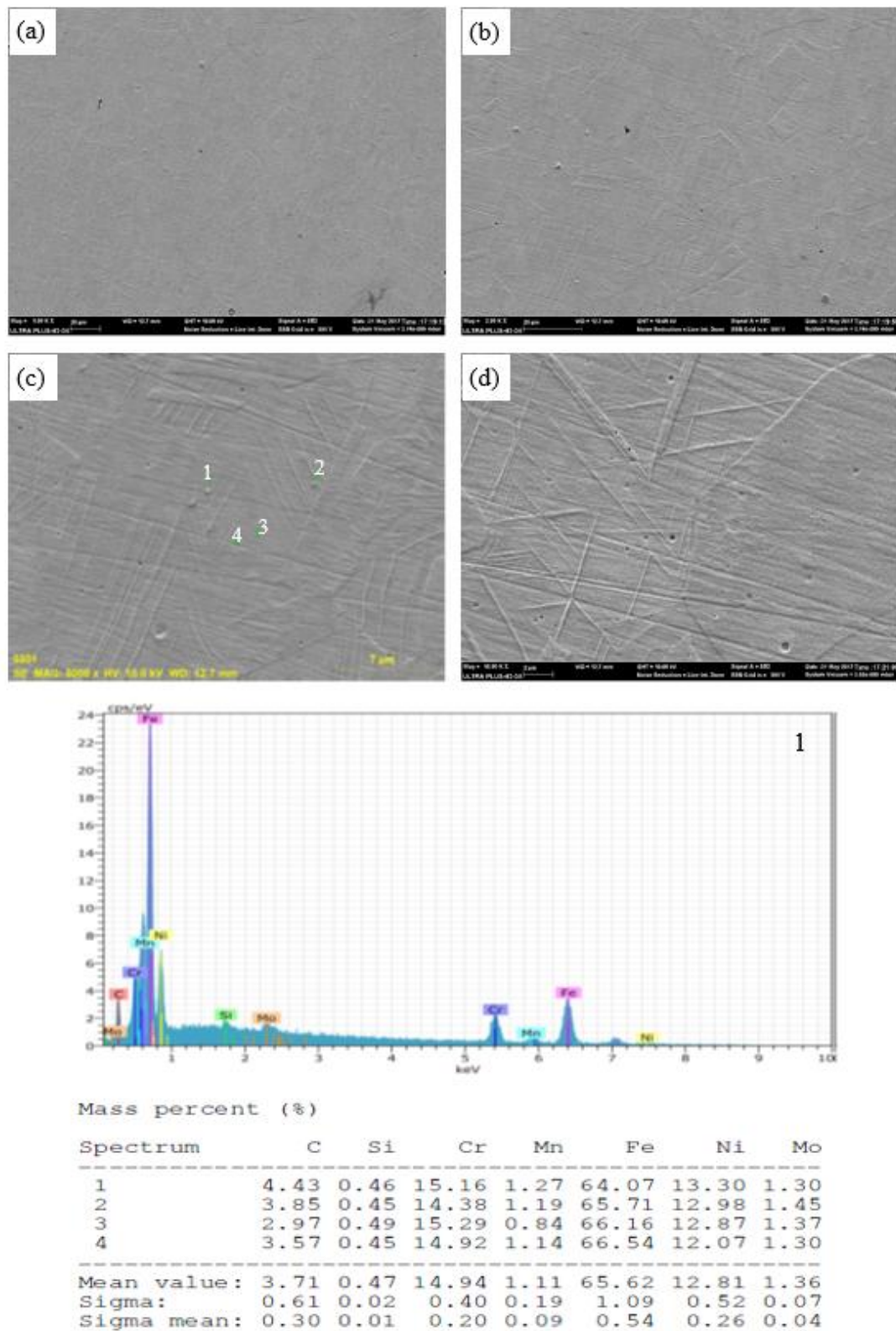
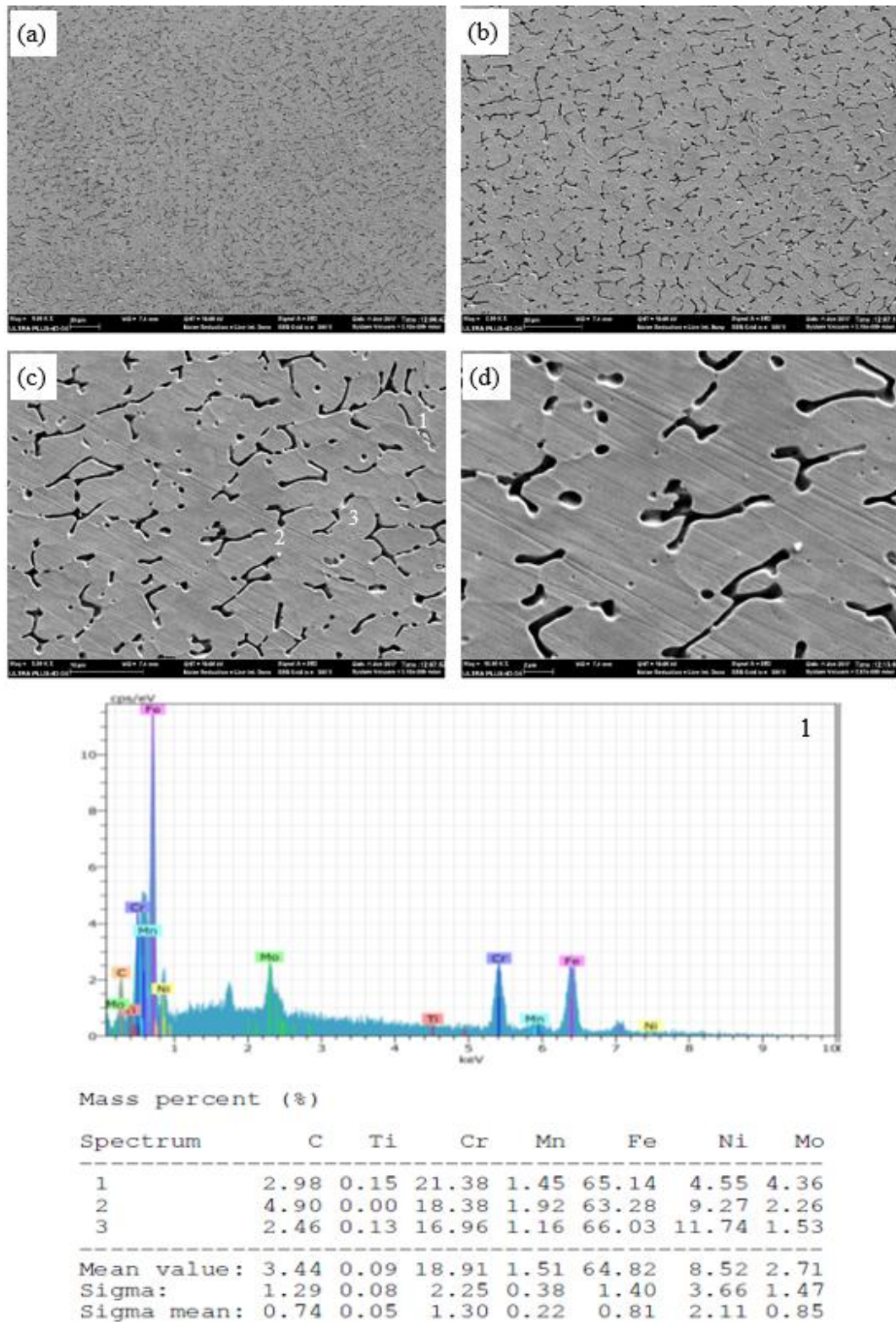


Figure 6.19 SEM micrographs with different magnification near fracture surface of as-welded samples tested at 25°C with EDS analyses.



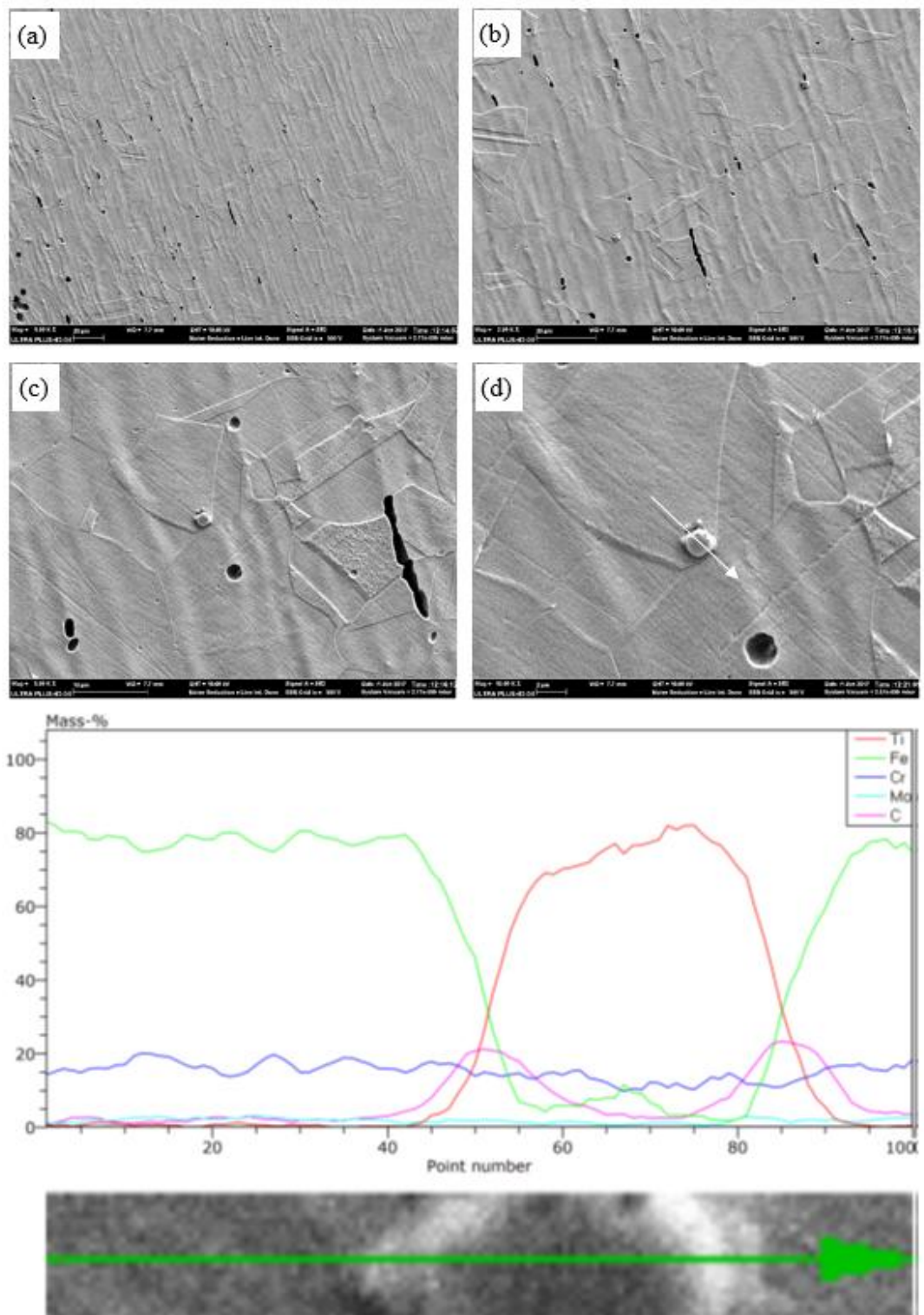


Figure 6.21 SEM micrographs with different magnification near fracture surface of as-welded samples tested at 400°C with line EDS analyses.

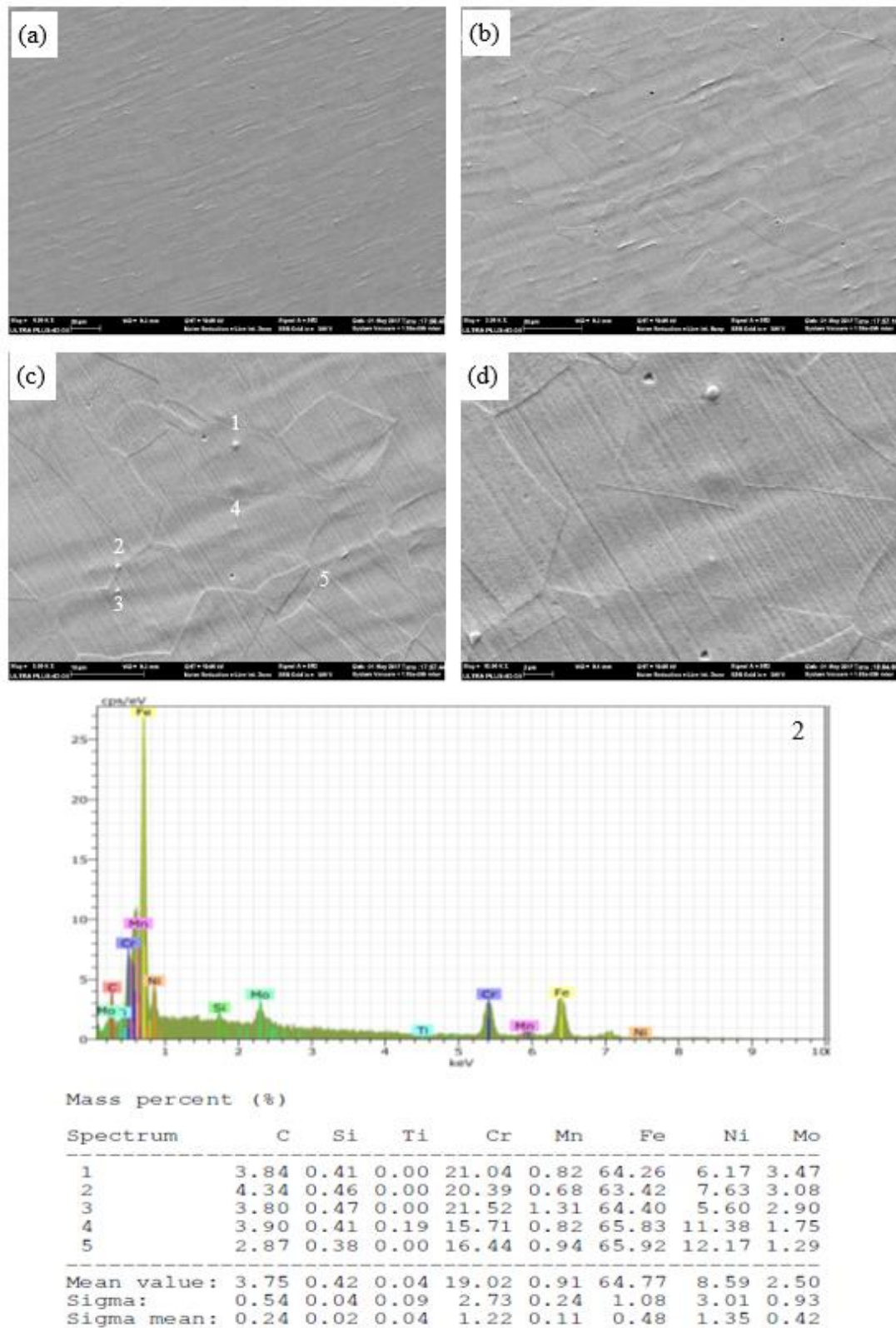


Figure 6.22 SEM micrographs with different magnification near fracture surface of as-welded samples tested at 600°C with EDS analyses.

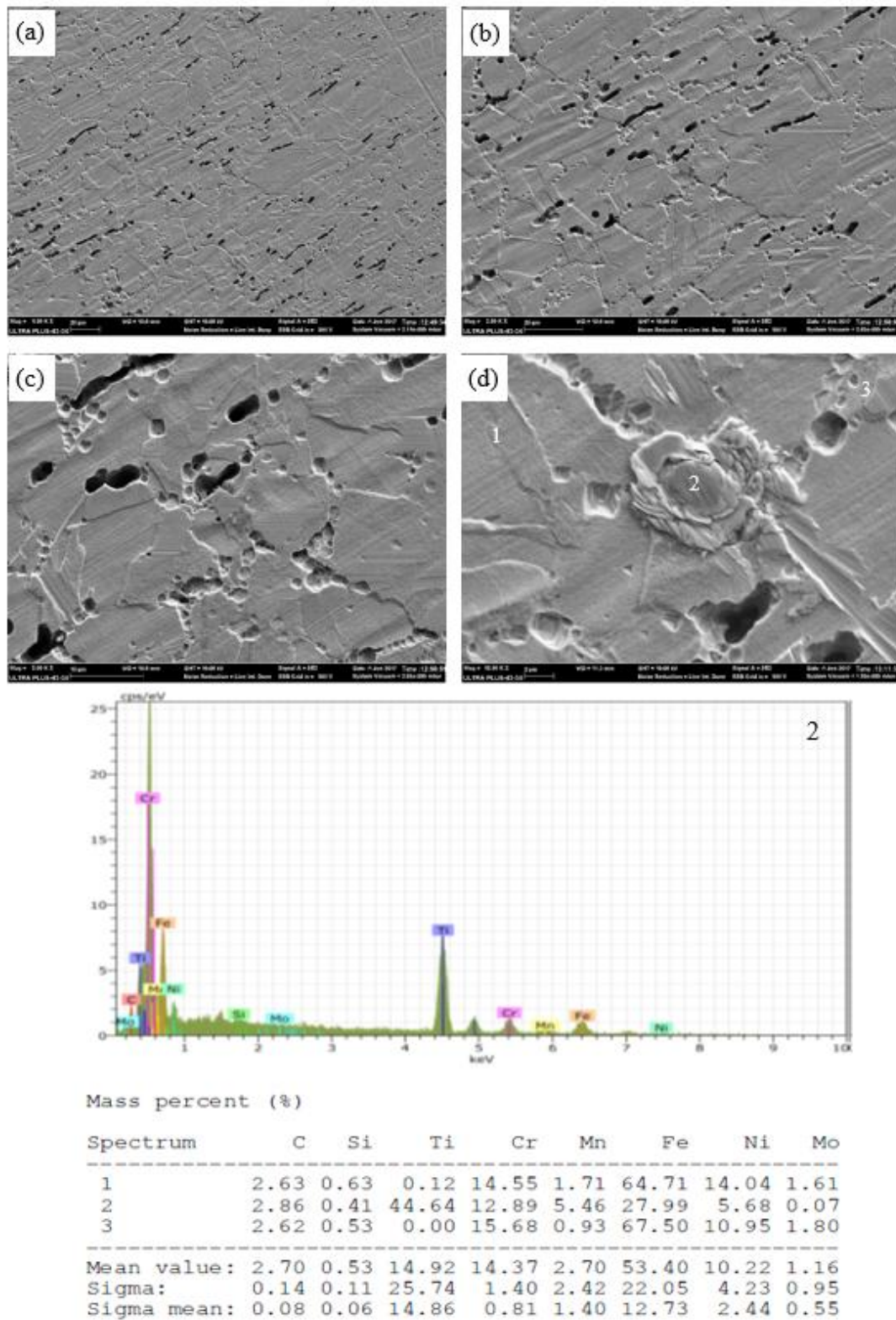


Figure 6.23 SEM micrographs with different magnification near fracture surface of as-welded samples tested at 800°C with EDS analyses.

6.4.1.3. Microstructure Analyses of Weld Metal Samples

Figures 6.24-6.28 show detailed the scanning electron micrographs and EDS analysis of the weld metal samples tensile tested at different temperatures. Figures 6.24, 6.25, 6.26, 6.27 and 6.28 show the microstructures near fracture surface of the weld metal samples tested at 25°C, 200°C, 400°C, 600°C and 800°C respectively. As it is seen, the weld metal consisted of skeletal ferrite in austenitic matrix with some carbides which were distributed in the austenite matrix (Figures 6.24-6.28). It was reported that the precipitation of carbides such as $M_{23}C_6$, MC, M_6C and M_7C_3 (M = Cr, Mo and Fe) occurs in austenitic stainless steel during heat treatment and welding. The precipitation of carbides which depends the temperature and carbon content reduces corrosion resistance of stainless steel. The temperature in the range from 550°C to 900°C allows chromium to diffuse away from the grain boundaries and to form carbides [55, 91-93]. Figures 6.24-6.28 also reveals EDS analysis with the different spectrum points marked on the microstructure of samples tested at 25-800°C. For example, Figure 6.26 shows EDS analysis with the points 1-4 on the samples tested at 200°C. Points 1-3 contain Mo, Cr and C which indicated that $Cr_{23}C_6$ and Mo_6C precipitate particles occurred in the weld metal during cooling after welding. Point 4 taken from ferrite phase also contains higher Cr.

The precipitation of the sigma phase, which is often observed in various series of stainless steels, is one of the main reasons for the deterioration of stainless steels' properties, for example, mechanical property, corrosion resistance, and weldability [94]. Sigma phase which is a tetragonal crystal structure forms in the Fe-Cr system at temperatures of 550-900°C. The presence of sigma phase in the matrix can increase the hardness and decrease the toughness, as well as the elongation of steel [95]. In the present work, the presence of sigma phase in the weld metal after testing in the temperature range of 100-800°C was not observed (Figures 6.24-6.28). It is thought that tensile testing time at elevated temperatures was not enough for precipitation of sigma phase. The sigma phase was a second phase of a rich substitutional element. The diffusion of the substitutional element was very slow in austenite so that the nucleation of the sigma phase in austenite was difficult [91].

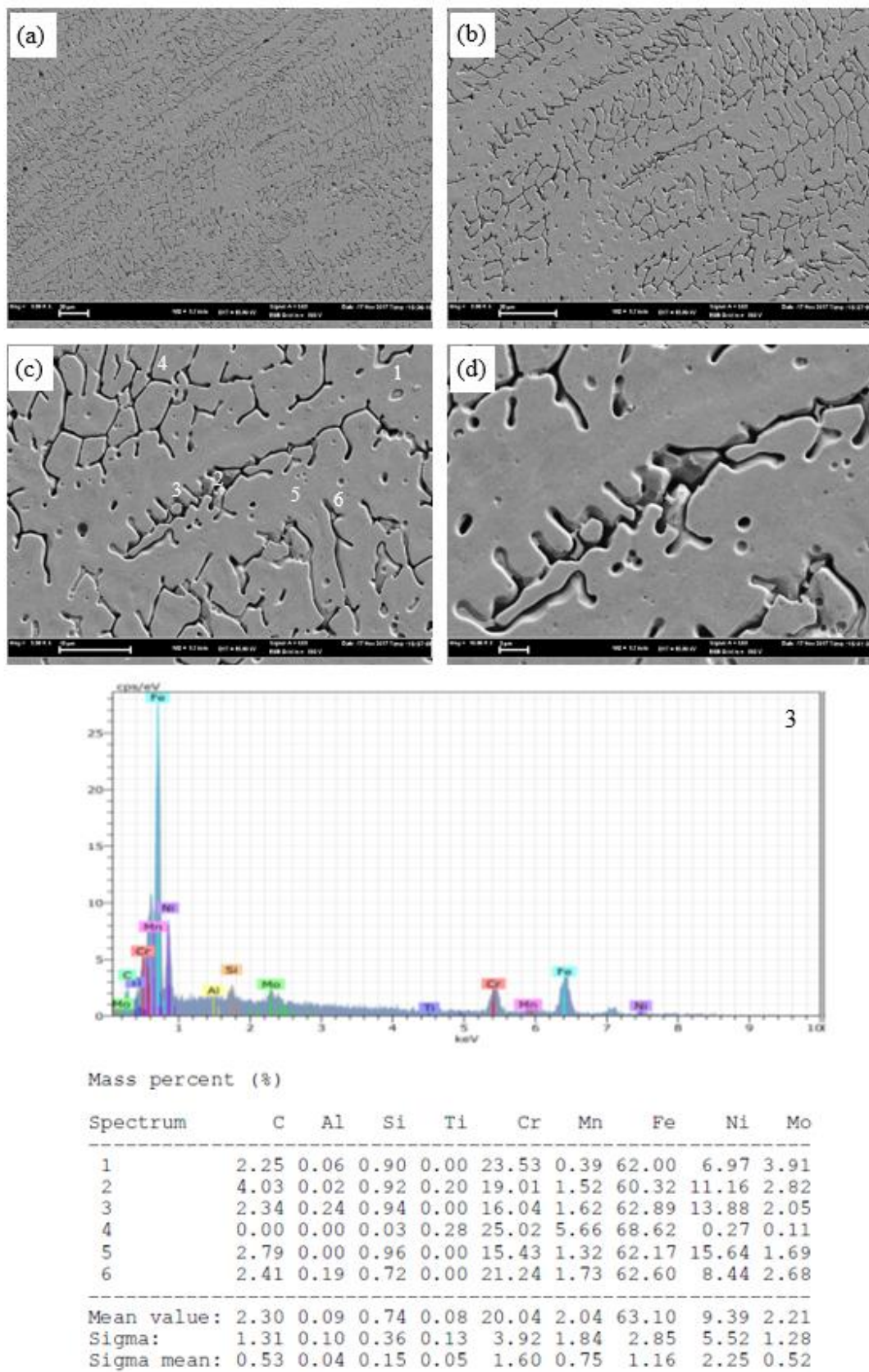
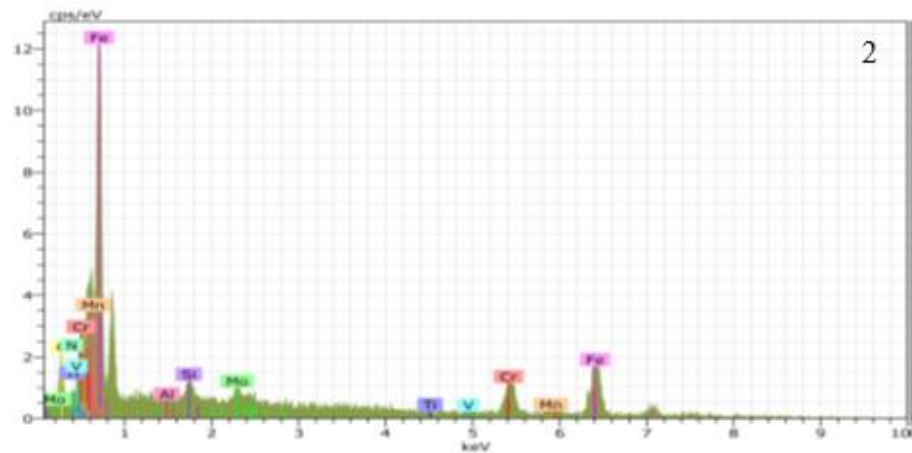
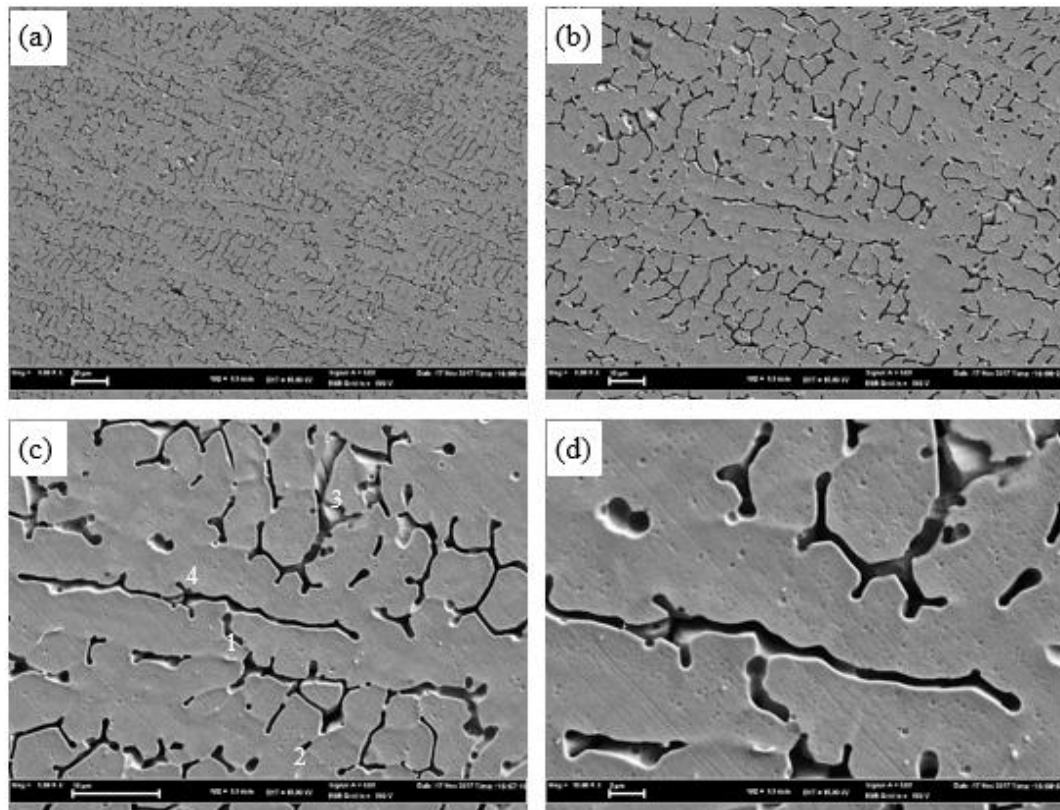


Figure 6.24 SEM micrographs with different magnification near fracture surface of weld metal samples tested at 25°C with EDS analyses.



Mass percent (%)

Spectrum	C	N	Al	Si	Ti	V	Cr	Mn	Fe	Mo
1	5.00	1.09	0.00	0.77	0.00	0.00	17.76	2.24	71.40	1.73
2	4.89	1.27	0.00	0.87	0.00	0.81	17.28	2.38	70.66	1.84
3	3.31	0.09	0.00	0.94	0.40	0.00	19.90	2.21	71.47	1.68
4	0.72	0.00	0.26	0.39	0.24	1.31	20.36	5.38	69.91	1.42
Mean value:	3.48	0.61	0.07	0.74	0.16	0.53	18.83	3.05	70.86	1.67
Sigma:	1.99	0.66	0.13	0.24	0.19	0.65	1.53	1.55	0.73	0.18
Sigma mean:	1.00	0.33	0.07	0.12	0.10	0.32	0.77	0.78	0.37	0.09

Figure 6.25 SEM micrographs with different magnification near fracture surface of weld metal samples tested at 200°C with EDS analyses.

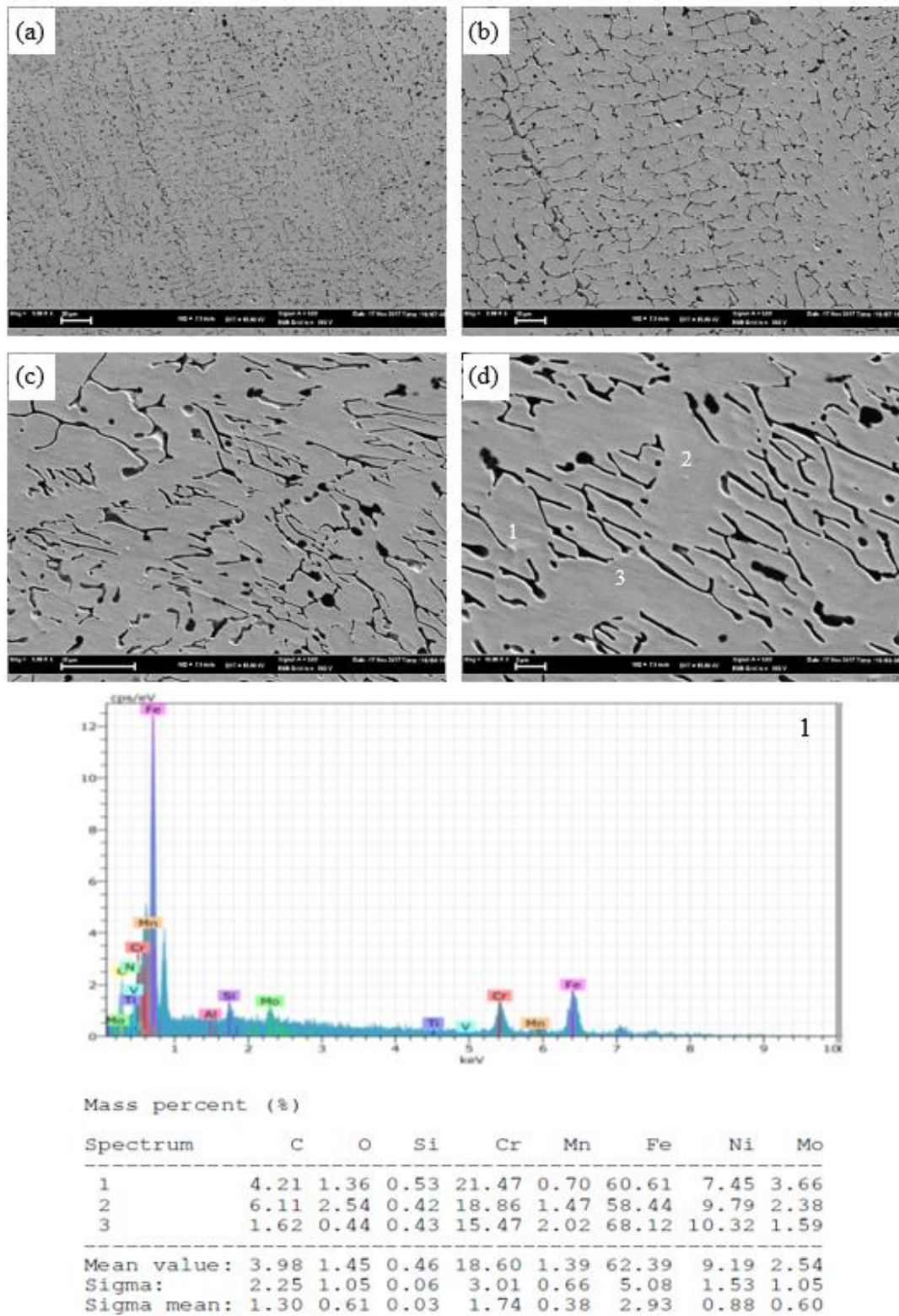


Figure 6.26 SEM micrographs with different magnification near fracture surface of weld metal samples tested at 400°C with EDS analyses.

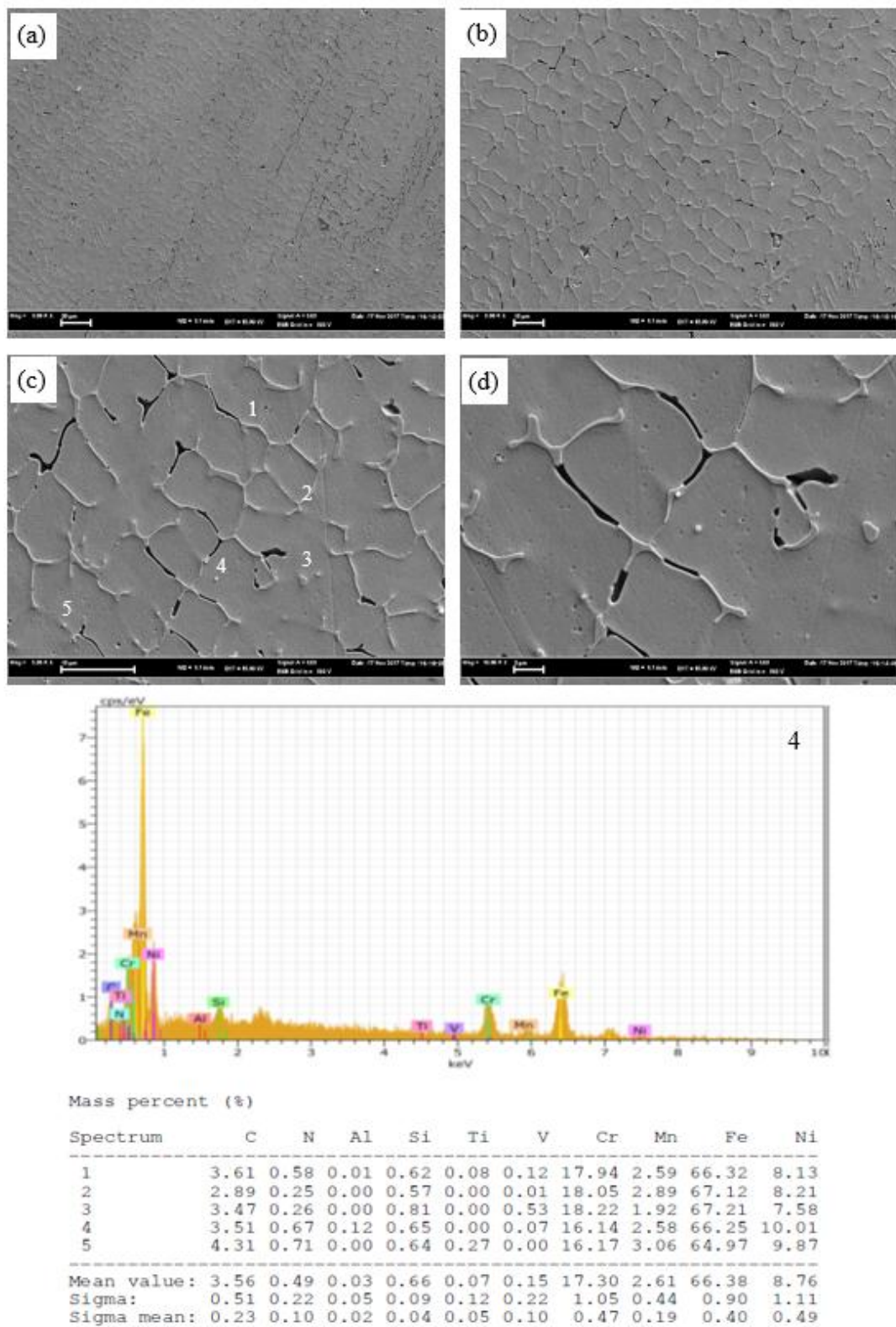


Figure 6.27 SEM micrographs with different magnification near fracture surface of weld metal samples tested at 600°C with EDS analyses.

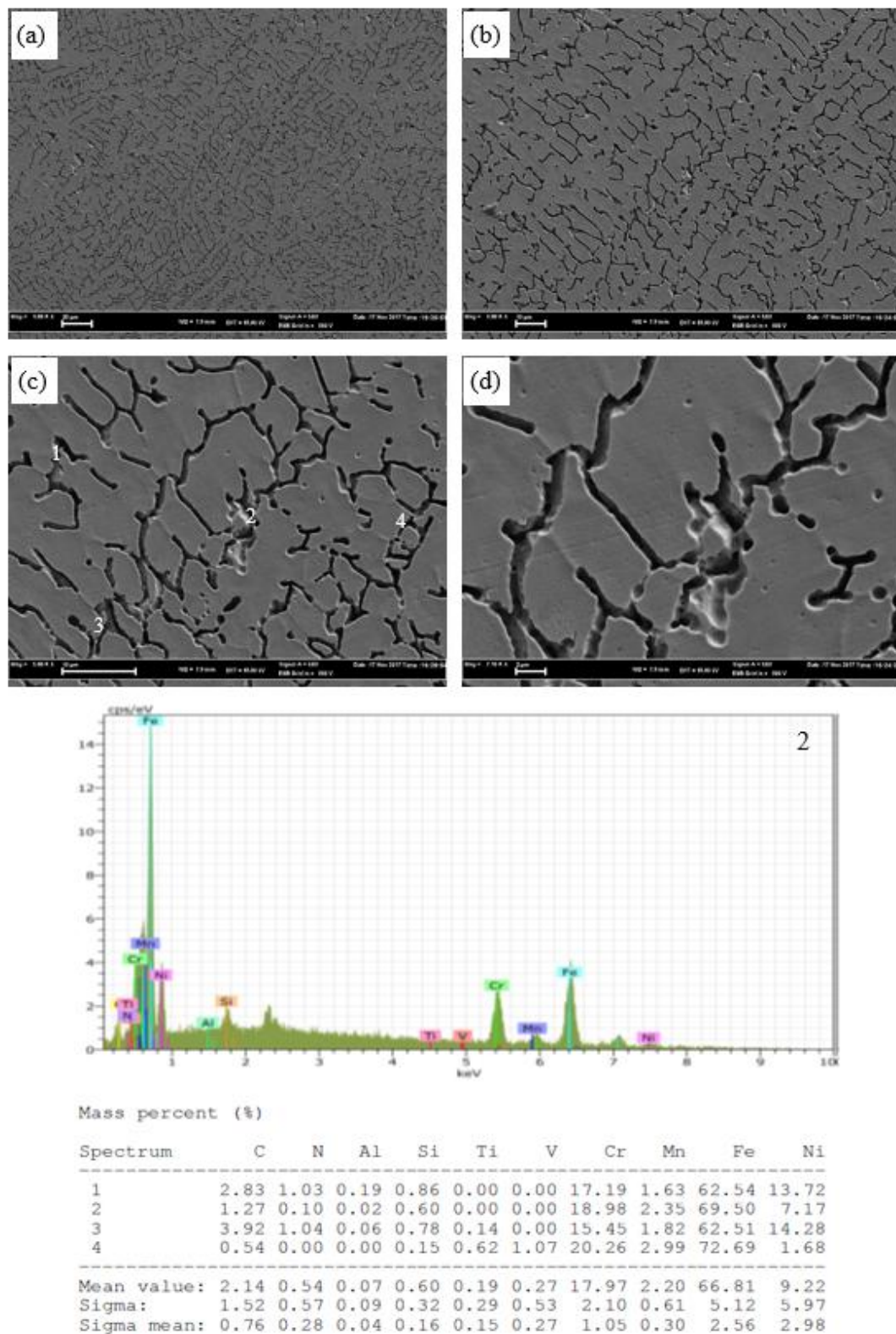


Figure 6.28 SEM micrographs with different magnification near fracture surface of weld metal samples tested at 800°C with EDS analyses.

6.4.2. Fracture Surface Analyses of As-received Samples, As-welded Samples and Weld Metal Samples.

6.4.2.1. Fracture Surface Analyses of As-Received Samples

To show the fracture type after the tensile test, the fractured specimens were examined using SEM. Figures 6.29-6.33 show fracture morphology of the as-received samples for the testing temperatures of 25–800°C. The fracture surface of as-received samples tensile tested at 25°C and 200°C showed ductile dimple fracture which is characteristic of ductile fracture. This indicates a ductile fracture mode with microvoid morphology, which is associated with the nucleation, growth and coalescence of microcavities [96].

At 400°C and 600°C, as-received samples revealed mixed type fracture of cleavage facets and dimples. The reduction in area is also decreased at the same temperatures of 400°C or 600°C, which suggests that dynamic strain ageing occurs in the 316L austenitic stainless steel under as-received condition. On the other hand, ductile dimples were found in as-received sample after testing at 800°C, which led to the increase in elongation and reduction in area.

Figures 6.29-6.33 also reveals EDS analysis with the different spectrum points marked on the microstructure of samples tested at 25-800°C. Some inclusions in small holes were seen on the microfractographs of as-received samples. The EDS analysis showed the presence of complex Mn-Cr-Ti-Al inclusions in as-received samples.

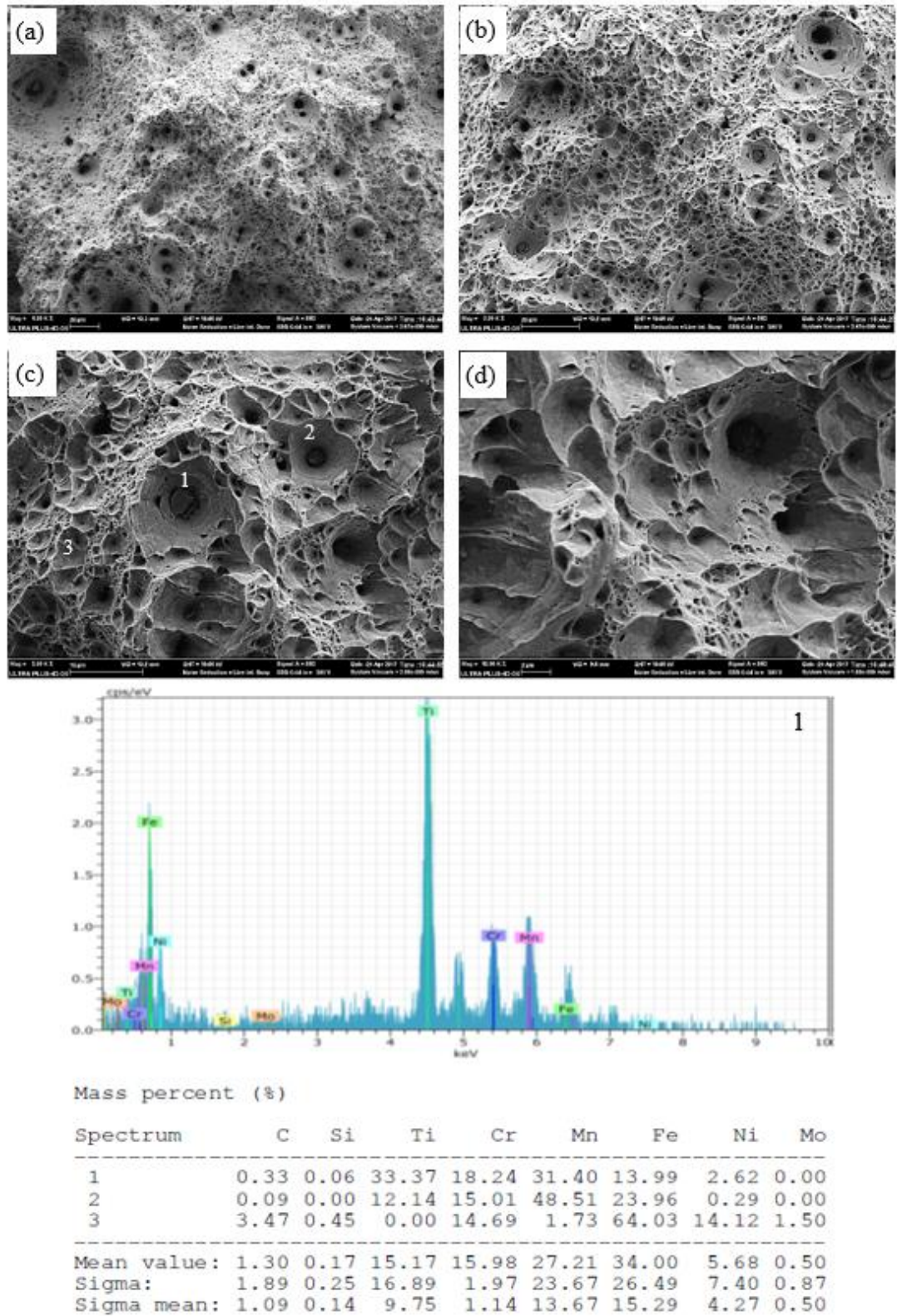


Figure 6.29 Fractography with different magnification of as-received samples tested at 25°C with correspondig EDS of the indicated particles.

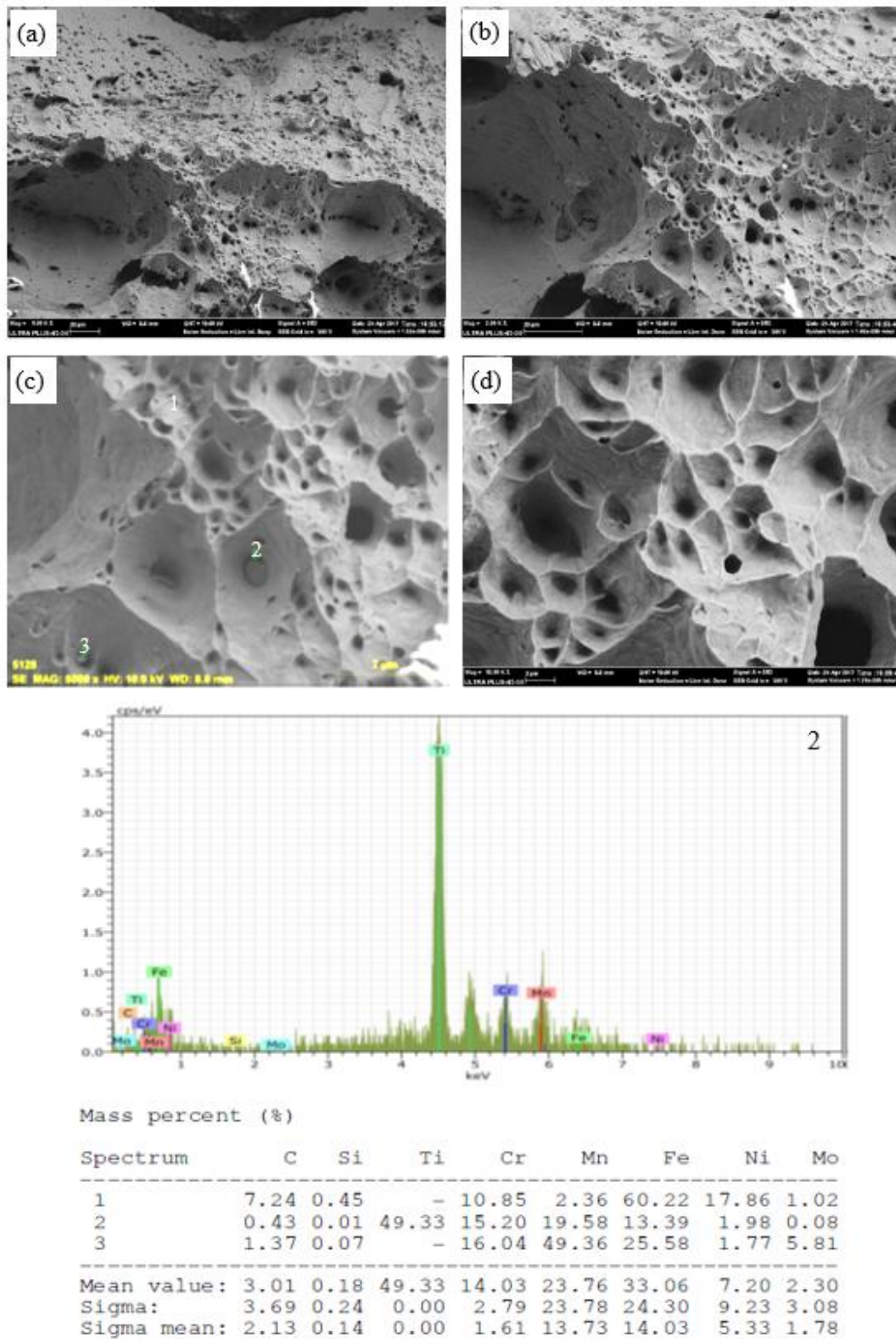


Figure 6.30 Fractography with different magnification of as-received samples tested at 200°C with correspondig EDS of the indicated particles.

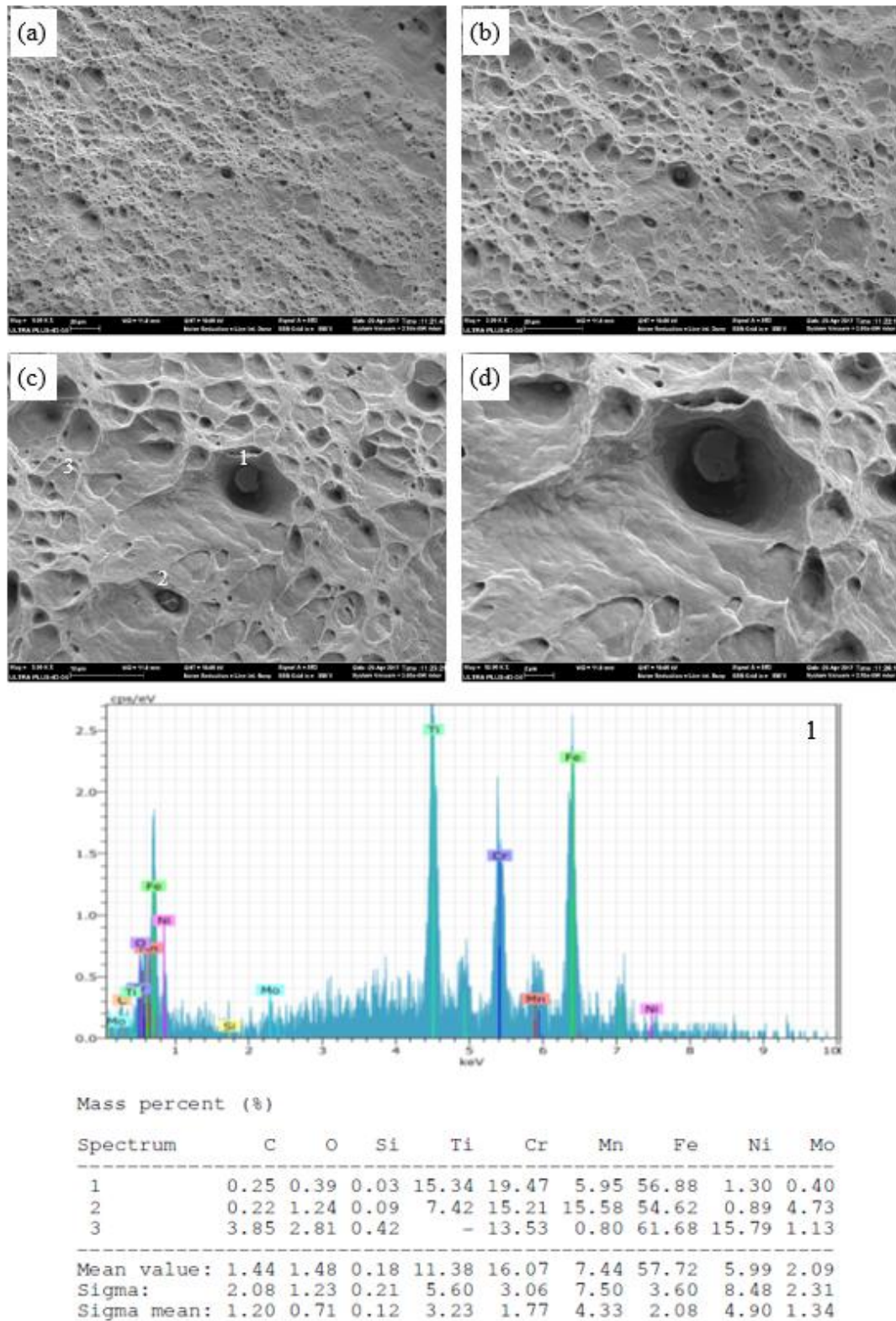


Figure 6.31 Fractography with different magnification of as-received samples tested at 400°C with correspondig EDS of the indicated particles.

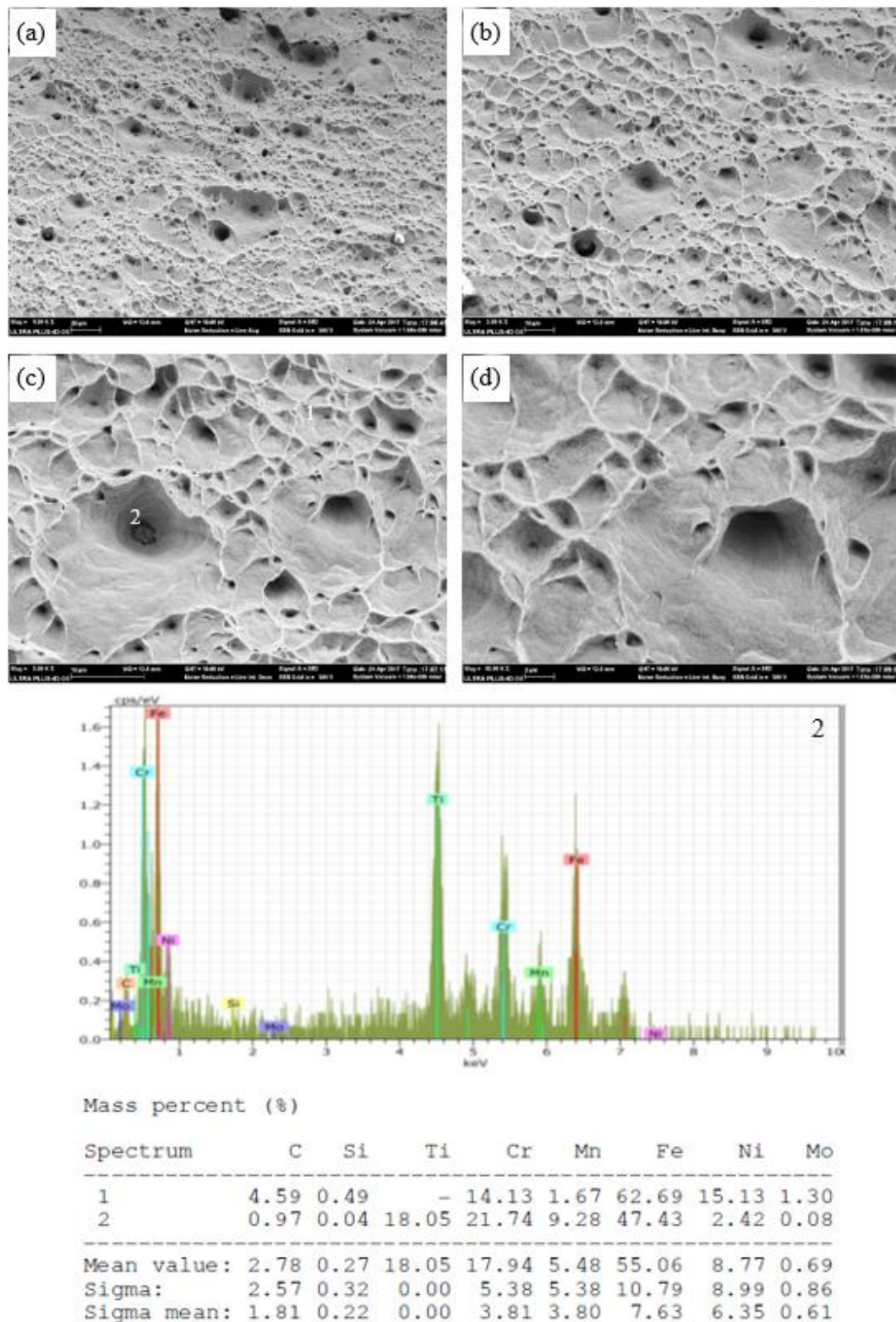


Figure 6.32 Fractography with different magnification of as-received samples tested at 600°C with correspondig EDS of the indicated particles.

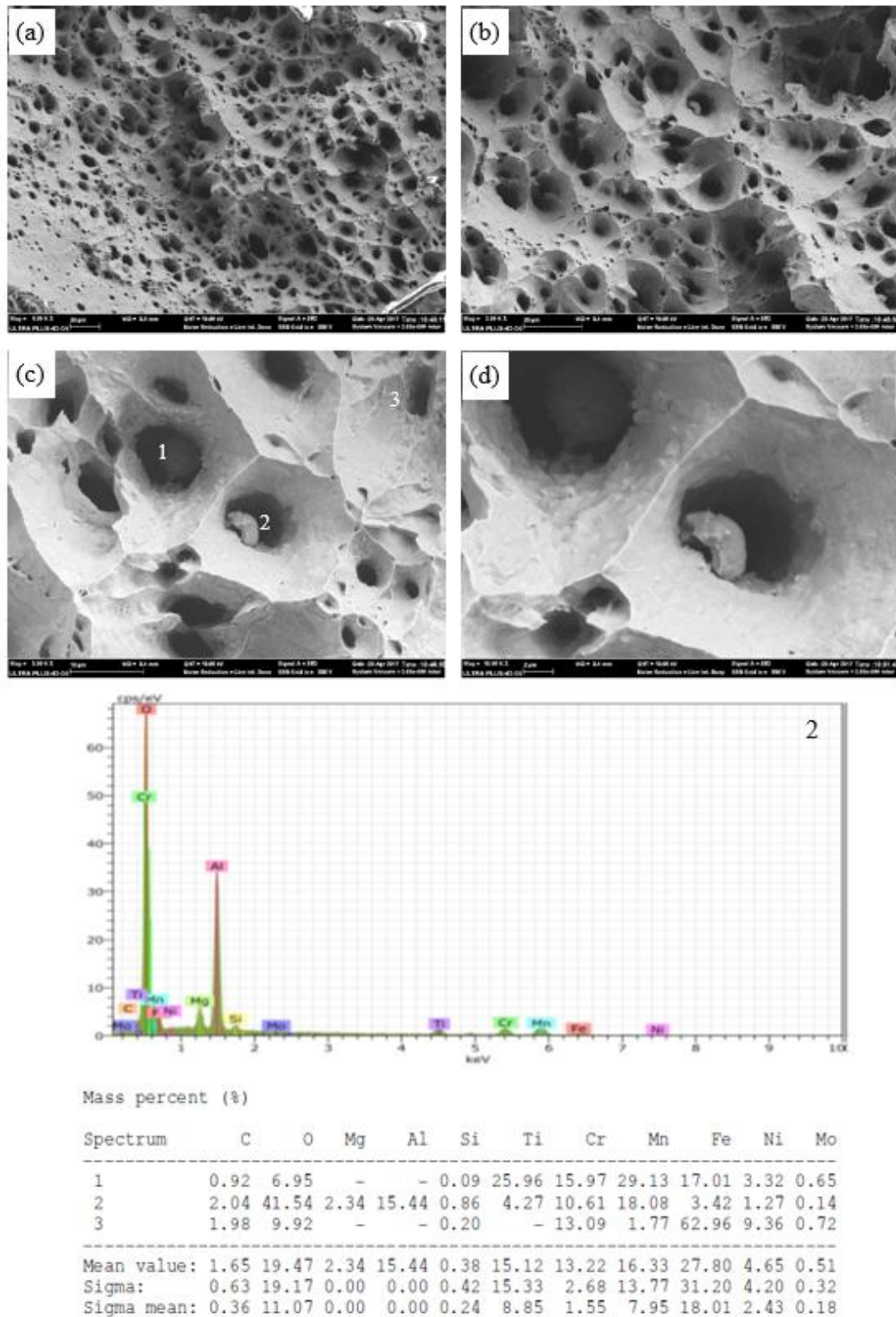


Figure 6.33 Fractography with different magnification of as-received samples tested at 800°C with correspondig EDS of the indicated particles.

6.4.2.2. Fracture Surface Analyses of As-Welded Samples

The fracture characteristics of tensile test samples were analyzed in the present experimental work. The macrographs of the as-welded samples are shown in Figure 6.34. The results revealed that the fracture formed in the HAZ of as-welded samples tested at 25–600°C. The fracture of the as-welded samples tested at temperatures of 25–600°C started from adjacent to the fusion zone in the HAZ and continued through into the weld metal. The strains developed during the thermal cycle of welding in the HAZ region is primarily responsible for starting fracture [97]. Due to strains next to fusion line can encourage to the starting failure from this region. However, the fracture in as-welded samples tested at 700°C and 800°C occurred in the base metal.



Figure 6.34 Failure mode of as-welded samples tensile tested at different temperatures.

In order to show the fracture type after the tensile test, the fractured specimens were examined by using SEM. Figures 6.35-6.39 show fracture morphology of the as-welded samples for the testing temperatures of 25-600°C. The fracture surface of as-welded samples tensile tested at 25°C and 200°C showed ductile dimple fracture which is characteristic of ductile fracture. This indicates a ductile fracture mode with microvoid morphology, which is associated with the nucleation, growth and coalescence of microcavities [98]. At 400°C and 600°C, as-welded samples revealed mixed type fracture of cleavage facets and dimples. The reduction in area is also decreased at the same temperatures of 400°C or 600°C which suggests that dynamic strain ageing occurs in the 316L austenitic stainless steel under as-welded conditions because of interaction between mobile dislocations and solute atoms.

However, as-welded samples showed lower decrease in area than as-received samples owing to dynamic strain ageing. On the other hand, ductile dimples were found in as-welded samples after testing at 800°C which led to the increase in elongation and reduction in area. Gündüz and Acarer [99] observed dimple and cleavage fracture pattern at 300°C for microalloyed medium carbon steel. They indicated that the primary cause of mixed fracture pattern is a change in the workhardening behaviour of the samples because of the interaction between mobile dislocations and solute atoms.

Figures 6.35-6.39 also reveals EDS analysis with the different spectrum points marked on the microstructure of samples tested at 25-800°C. Some inclusions in small holes were seen on the microfractographs of as-received samples. The EDS analysis showed the presence of complex Mn-Cr-Ti-Al inclusions in as-welded samples.

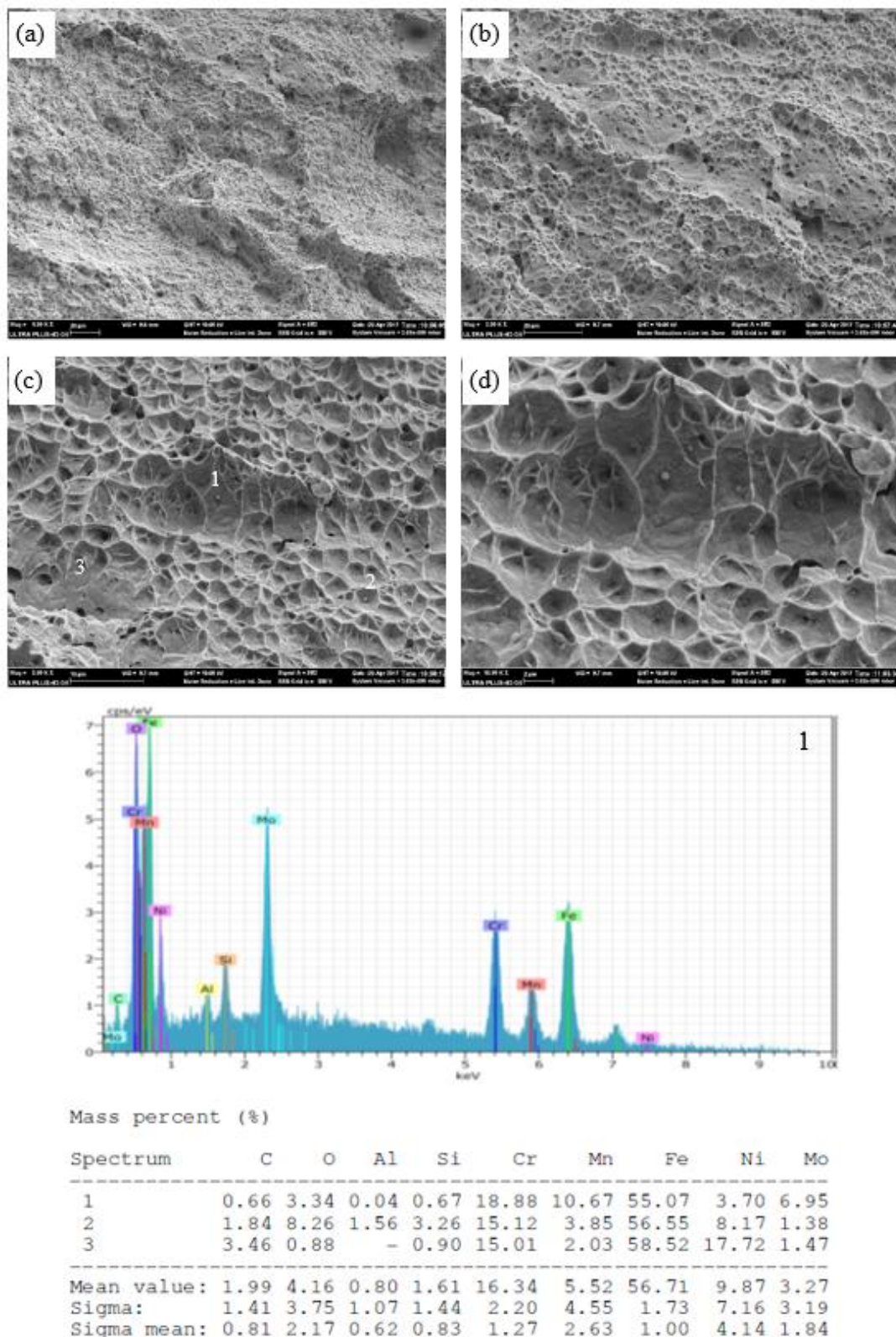


Figure 6.35 Fractography with different magnification of as-welded samples tested at 25°C with correspondig EDS of the indicated particles.

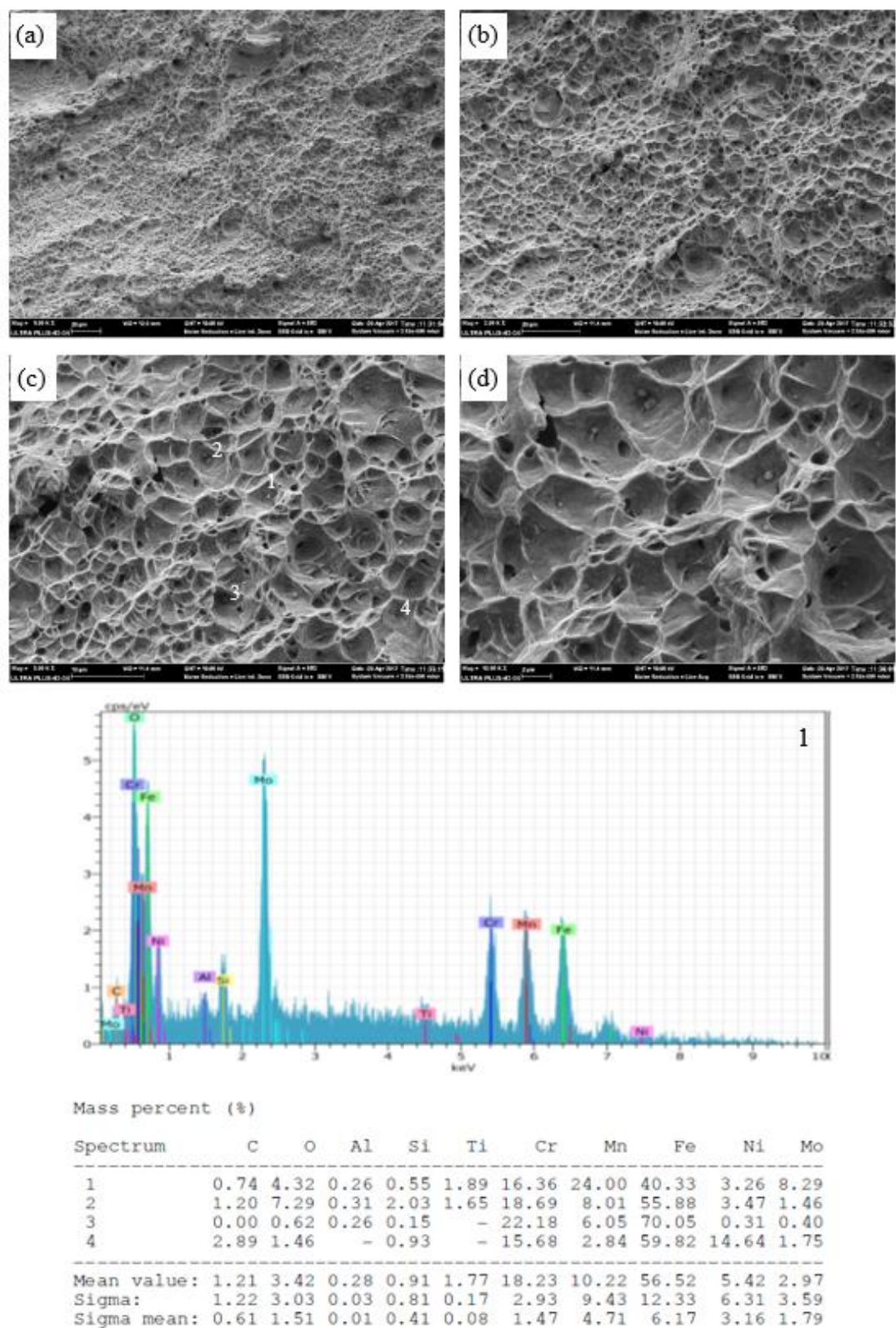


Figure 6.36 Fractography with different magnification of as-welded samples tested at 200°C with correspondig EDS of the indicated particles.

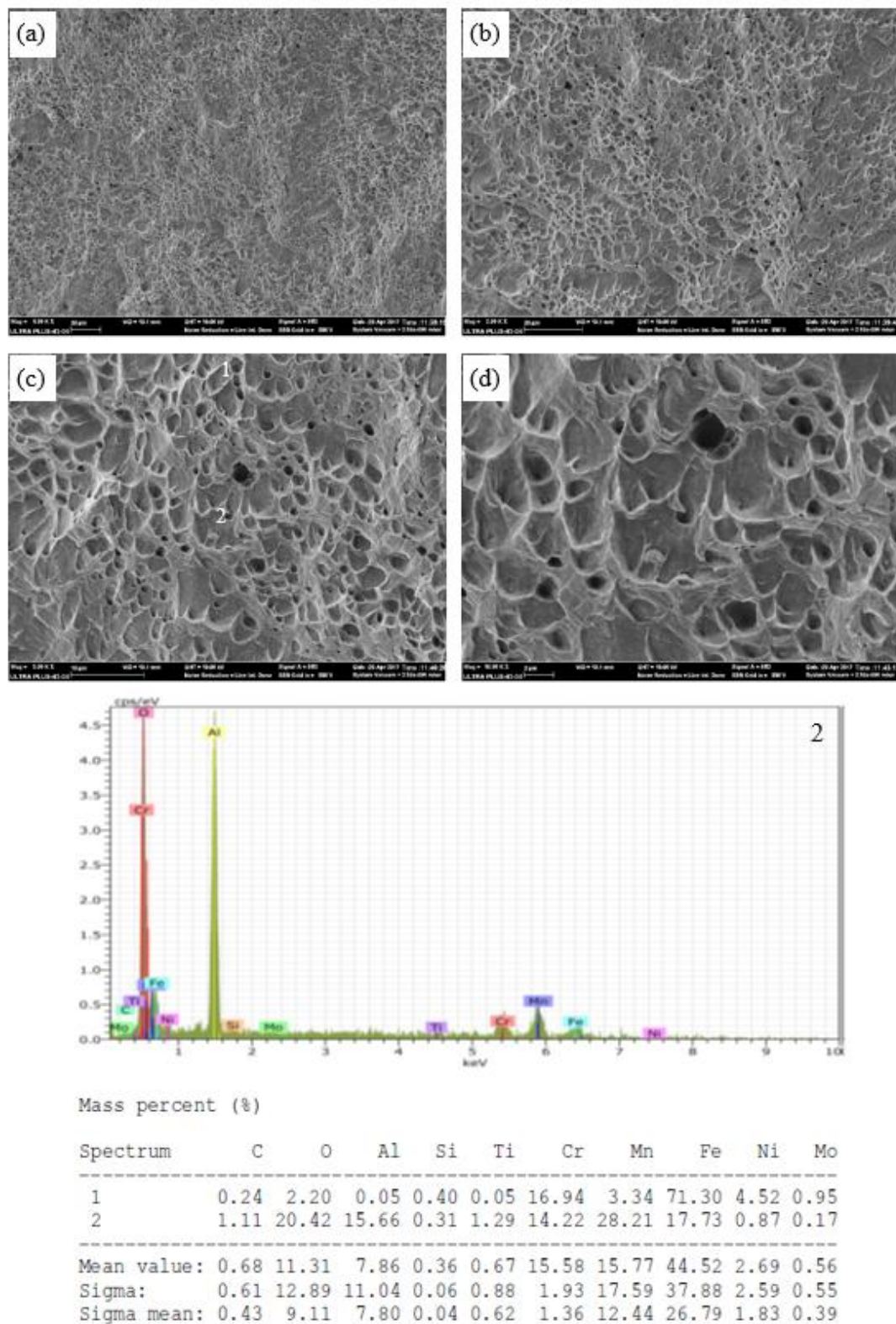


Figure 6.37 Fractography with different magnification of as-welded samples tested at 400°C with correspondig EDS of the indicated particles.

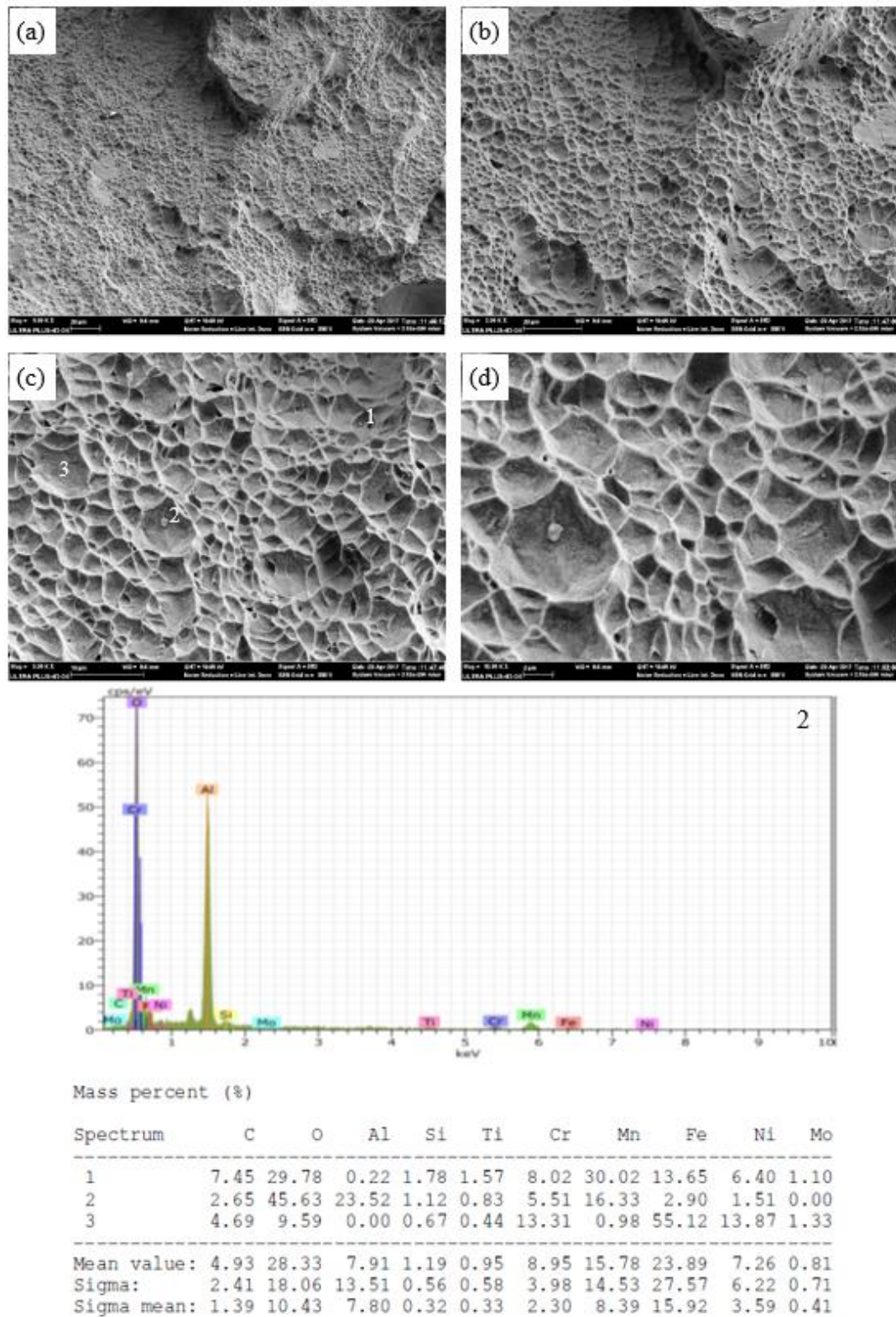
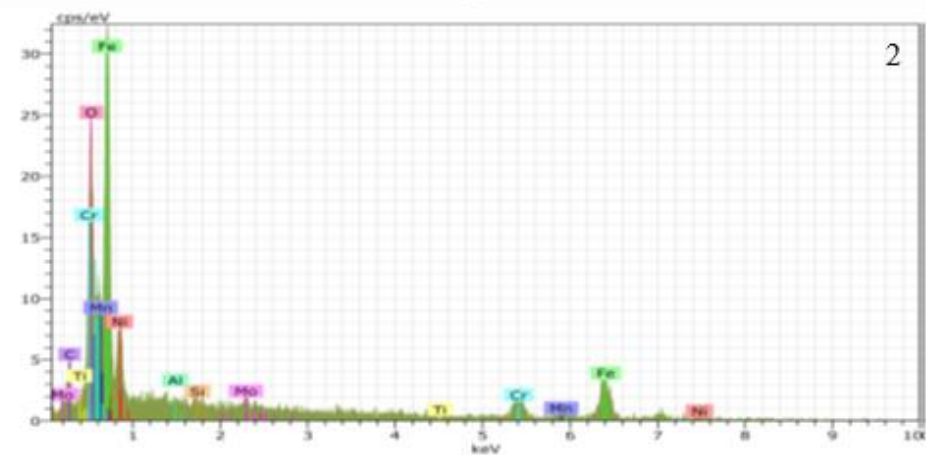
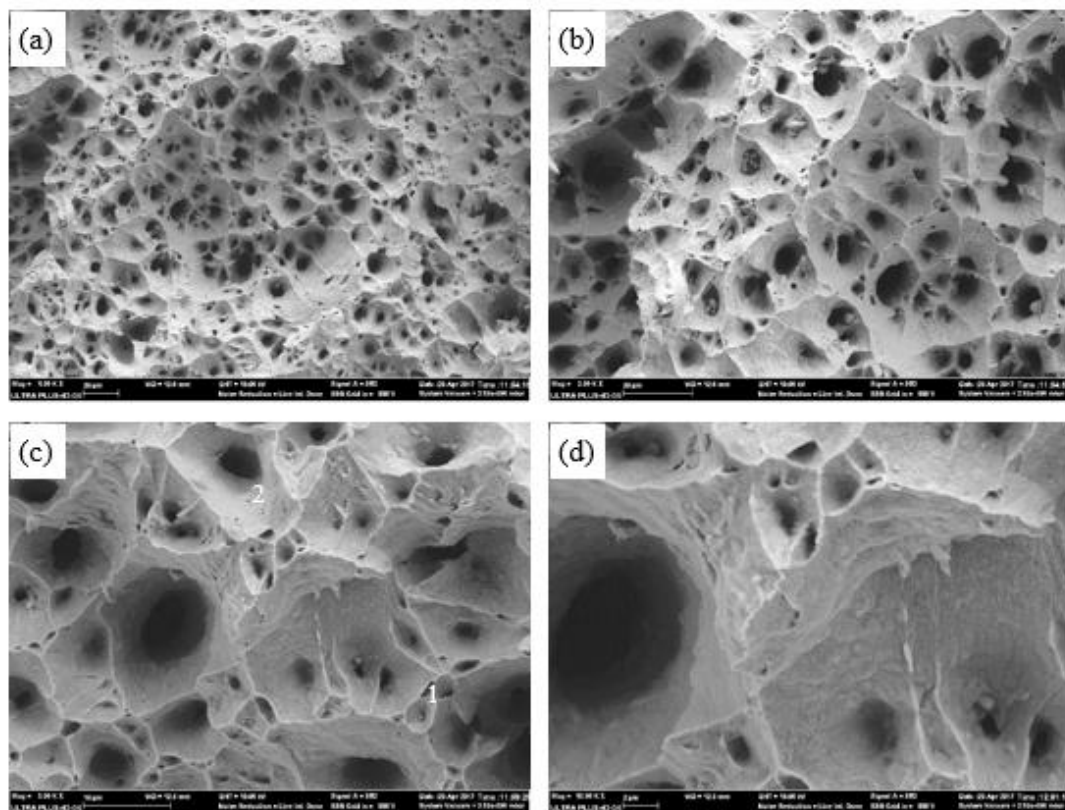


Figure 6.38 Fractography with different magnification of as-welded samples tested at 600°C with correspondig EDS of the indicated particles.



Mass percent (%)

Spectrum	C	O	Al	Si	Ti	Cr	Mn	Fe	Ni	Mo
1	0.29	0.49	0.00	0.01	0.00	10.35	55.21	33.29	0.15	0.21
2	4.86	12.21	0.19	0.30	0.00	11.72	0.92	56.25	12.67	0.89
Mean value:	2.58	6.35	0.09	0.16	0.00	11.04	28.06	44.77	6.41	0.55
Sigma:	3.23	8.29	0.13	0.20	0.00	0.97	38.39	16.24	8.85	0.48
Sigma mean:	2.28	5.86	0.09	0.14	0.00	0.68	27.14	11.48	6.26	0.34

Figure 6.39 Fractography with different magnification of as-welded samples tested at 800°C with correspondig EDS of the indicated particles.

6.4.2.3. Fracture Surface Analyses of Weld Metal Samples

To show the fracture type after the tensile test, the fractured specimens were examined using SEM. Figures 6.40-6.44 show fracture morphology of the weld metal samples for the testing temperatures of 25–800°C. The fracture surface of weld metal tensile tested at 25°C and 200°C showed dimples with different sizes which is characteristic of ductile fracture. Dimple morphology dominated on the fracture surface. This indicated that ductile fracture mode which was represented by cup-like dimple rupture occurred in weld metal [100, 101].

Cleavage facets and dimples were observed in the weld metal after testing at 400°C or 600°C in which the samples also showed lower decrease in reduction of area. This suggests that dynamic strain ageing occurs in weld metal of 316L austenitic stainless steel. On the other hand, ductile dimples were observed in weld metal after testing at 800°C which increased elongation and reduction in area.

Figures 6.40-6.44 also reveals EDS analysis with the different spectrum points marked on the microstructure of weld metal samples tested at 25-800°C. Some inclusions in small holes were seen on the microfractographs of weld metal samples. The EDS analysis showed the presence of complex Mn-Cr-Ti-Al inclusions in weld metal samples.

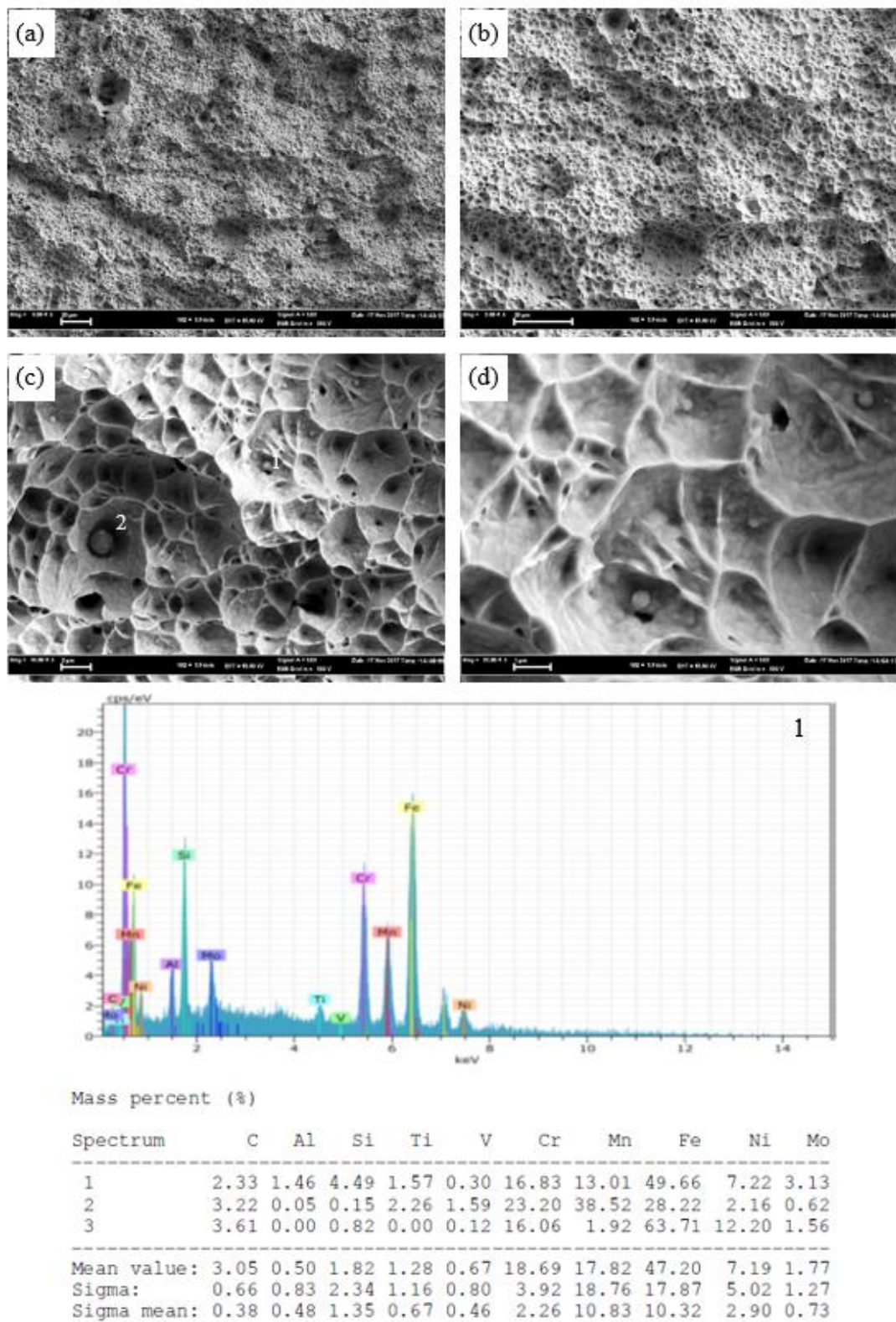


Figure 6.40 Fractography with different magnification of weld metal samples tested at 25°C with correspondig EDS of the indicated particles.

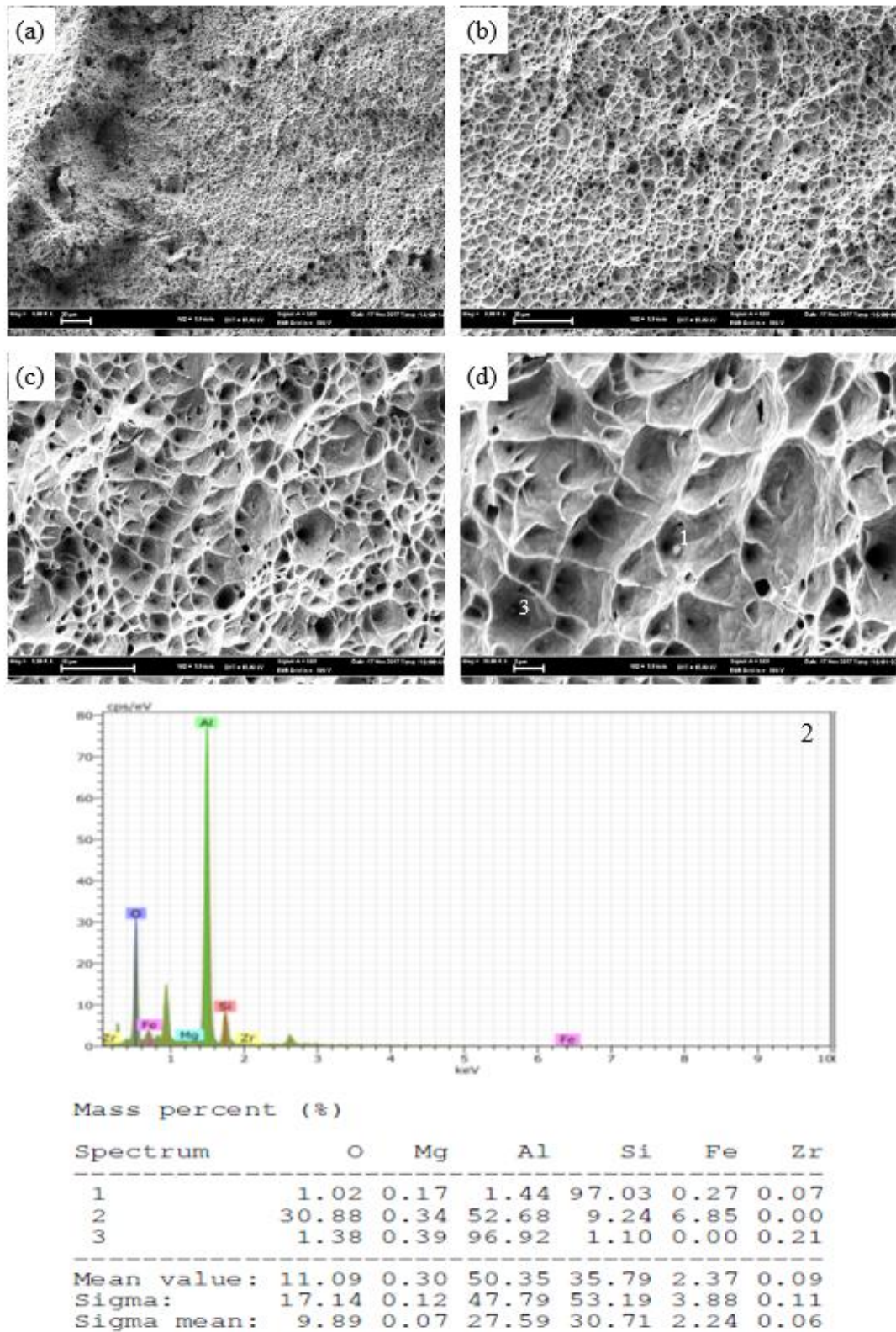


Figure 6.41 Fractography with different magnification of weld metal samples tested at 200°C with correspondig EDS of the indicated particles.

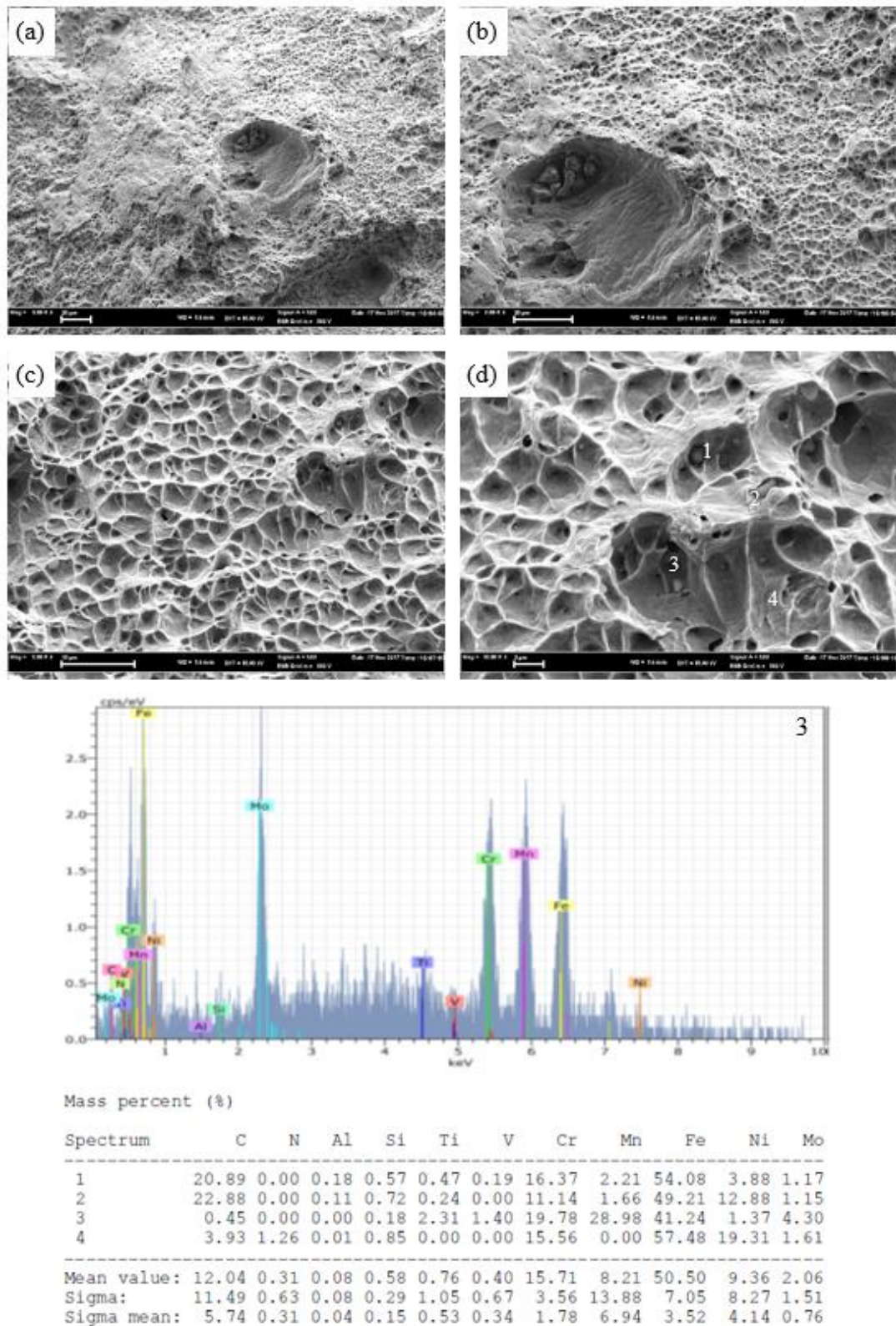


Figure 6.42 Fractography with different magnification of weld metal samples tested at 400°C with correspondig EDS of the indicated particles.

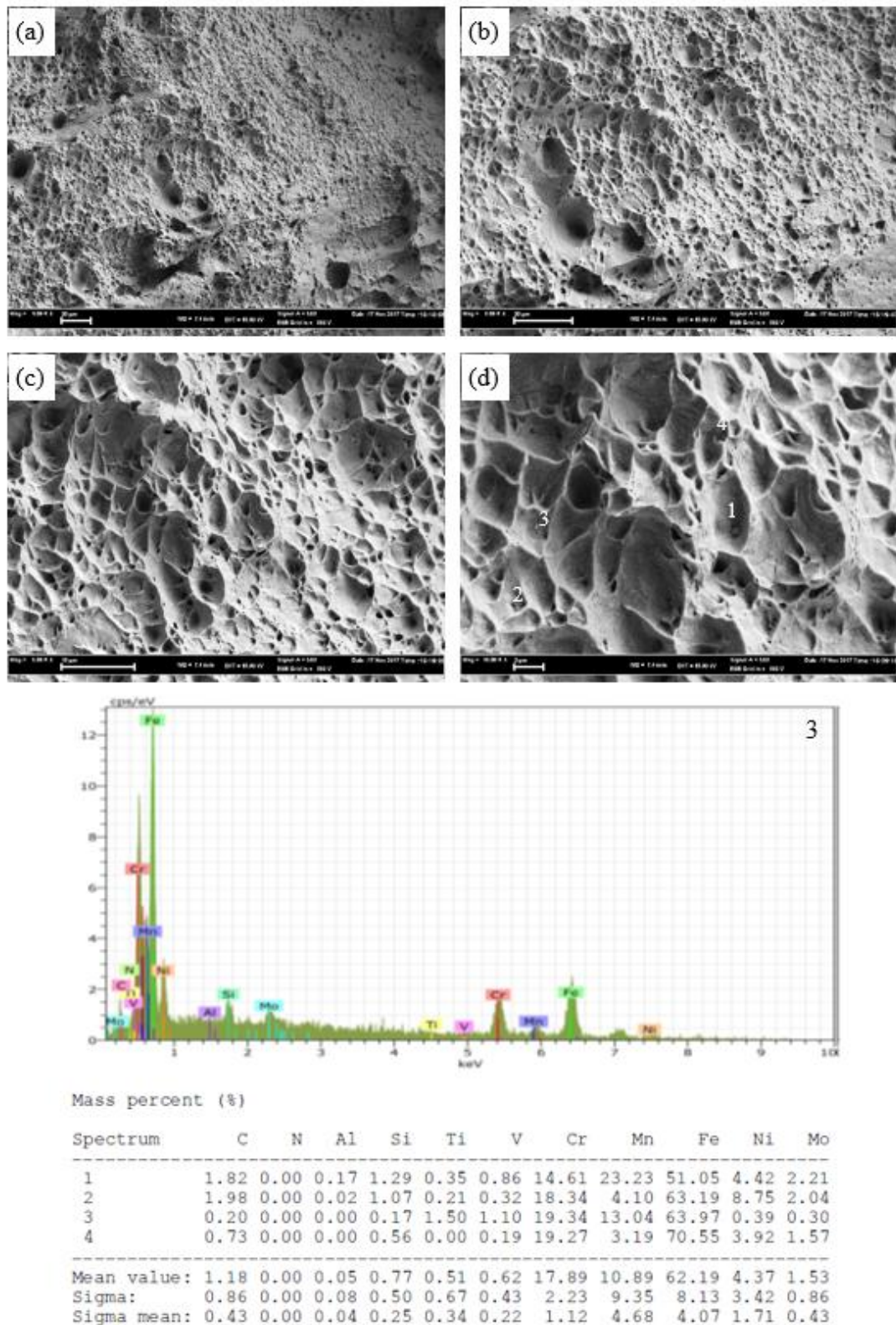
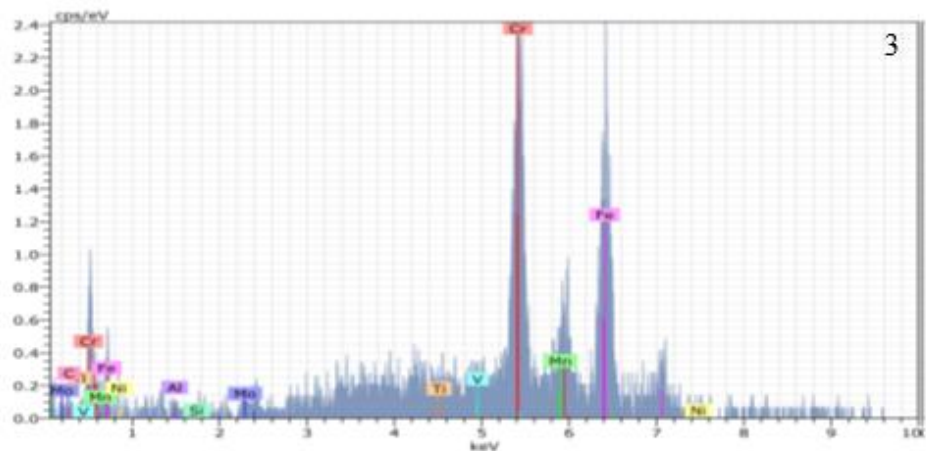
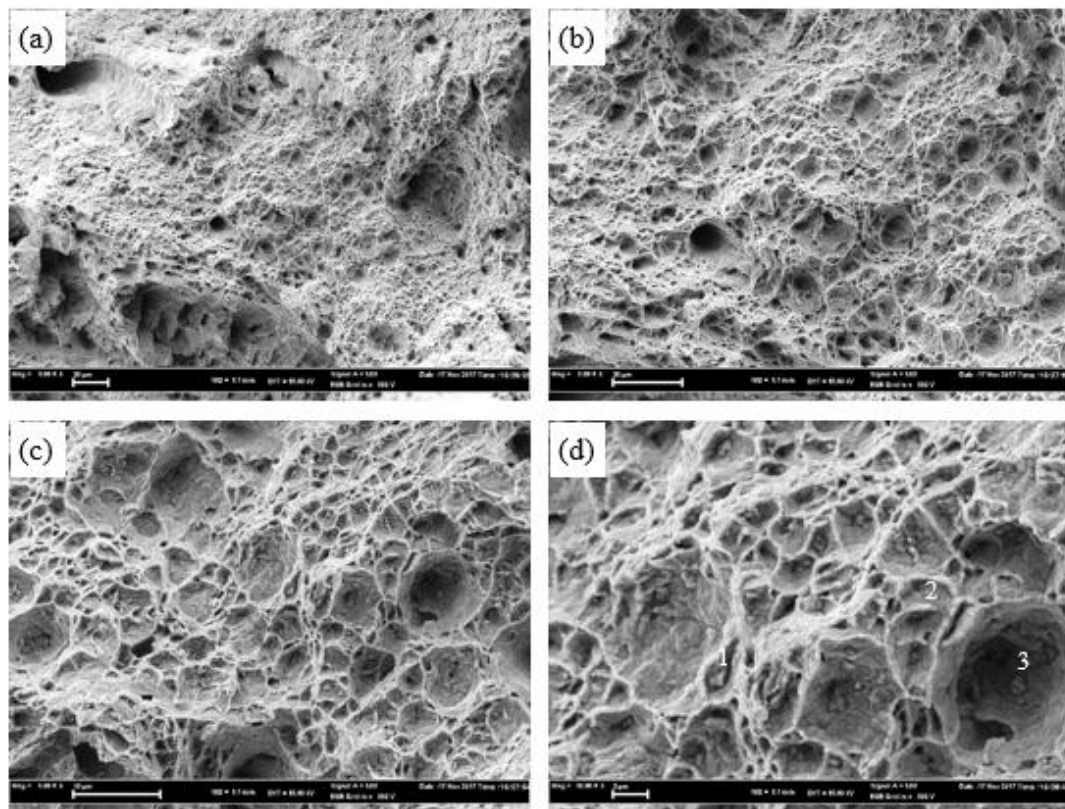


Figure 6.43 Fractography with different magnification of weld metal samples tested at 600°C with correspondig EDS of the indicated particles.



Mass percent (%)										
Spectrum	C	Al	Si	Ti	V	Cr	Mn	Fe	Ni	Mo
1	3.45	0.01	1.33	0.00	0.46	30.07	4.26	50.05	6.12	4.25
2	6.33	0.09	1.16	0.00	0.21	23.16	1.97	53.22	12.79	1.08
3	0.10	0.07	0.00	1.27	2.44	32.15	6.76	56.96	0.00	0.25
Mean value:	3.29	0.06	0.83	0.42	1.03	28.46	4.33	53.41	6.30	1.86
Sigma:	3.11	0.04	0.72	0.74	1.22	4.71	2.40	3.46	6.40	2.11
Sigma mean:	1.80	0.02	0.42	0.42	0.70	2.72	1.38	2.00	3.69	1.22

Figure 6.44 Fractography with different magnification of weld metal samples tested at 800°C with correspondig EDS of the indicated particles

CHAPTER 7

CONCLUSIONS AND SUGGESTION FOR FURTHER WORK

7.1. CONCLUSION

In the present work, the dynamic strain ageing behaviour of AISI 316L grade austenitic stainless steel was investigated under as-received, as-welded and weld metal conditions. Tensile tests were carried out at temperatures of 25-800°C at a strain rate of $1 \times 10^{-3} \text{s}^{-1}$. The conclusions obtained from present study can be given as follows:

1. As-received samples, as-welded samples and weld metal of 316L austenitic stainless steel which is susceptible to dynamic strain ageing showed small variation in YS and UTS values up to 600°C, but it revealed drastic change when the test temperature was increased beyond 600°C.
2. A decrease in elongation (%) values of as-received samples, as-welded samples and weld metal was observed with increasing temperature, reaching a minimum at temperature of 300-600°C. This is due to the the presence of free or uncombined C, N or Cr solute atoms in solution.
3. Weld metal and as-welded samples are more susceptible to dynamic strain ageing compared to the as-received samples, due to the partial dissolution of carbides in the weld metal and HAZ next to the fusion line during welding. Therefore, as-welded samples and weld metal revealed higher increase in YS and UTS values up to 600°C compared to the as-received samples.
4. Weld metal is more susceptible to dynamic strain ageing than the as-received samples and as-welded samples with evidence more increase in strength but decrease in elongation (%) at the testing temperatures of 25-800°C.

This is due to the presence of higher Cr content in weld metal compared to that of as-received samples and as-welded samples, because weld wire Cr content is about 18.5 wt. % which is higher than as-received samples.

5. Vickers microhardness measurement of the base metal, HAZ and weld metal showed that the highest hardness was obtained in the weld metal. This could be the presence of higher amount of Cr in solid solution due to welding thermal cycle. Higher hardness in weld metal may also be owing to melting and solidification of the area during or cooling after welding. Generally, as-welded samples revealed the lowest and highest hardness in the base metal and weld metal respectively. Hardness of HAZ was found to be higher than base metal, but lower than weld metal.
6. As-received samples of 316L austenitic stainless steel achieved higher workhardening index (n) values than as-welded samples and weld metal due to good formability. The as-received samples reached a larger reduction in workhardening index (5%) while as-welded samples showed an increase in workhardening index (23%) at the peak ageing temperature of 500°C, compared to those in the room temperature testing conditions.
7. Dynamic strain ageing is more pronounced in weld metal samples compared to the as-received samples and as-welded samples. It was found that weld metal showed the lowest workhardening index for all testing temperatures compared to the as-received samples and as-welded samples due to poor formability which is resulted in dynamic strain ageing.
8. 316L austenitic stainless steel has high susceptibility to deformation induced martensite after testing at 25°C. The results obtained from present study indicated that the martensite occurred at the shear bands.
9. Stress and strain diagrams of as-received samples, as-welded samples and weld metal samples revealed serrated behaviour when the testing temperatures increased at a strain rate of $1 \times 10^{-3} \text{ s}^{-1}$. Serrated behavior which

is the characteristic of dynamic strain ageing was observed during tensile testing at 500°C and 600°C. The serrations were not observed on the stress and strain diagrams after testing at 700°C or 800°C.

10. Fracture surface analysis indicated that as-received samples, as-welded samples and weld metal samples showed ductile fracture mode after testing at 25°C, 200°C and 800°C. However, as-received samples, as-welded samples and weld metal revealed mixed type fracture of cleavage facets and dimples at 400°C and 600°C. The reduction in area is also decreased at the same temperatures of 400°C or 600°C, which suggests that fracture occurs more easily under dynamic strain ageing conditions.

7.2. SUGGESTION FOR FURTHER WORK

The conclusions drawn above indicate that the high temperature tensile behaviour of 316L austenitic stainless steel under as-received, as-welded and weld metal conditions depends on the presence of solute atoms such as C, N, Cr and Mo. However, further work would be required in order to confirm the trends observed through this work. For example in the present experimental work, dynamic strain ageing test was carried out for the as-received samples, as-welded samples and weld metal tested at 25-800°C for constant strain rate of $1 \times 10^{-3} \text{ s}^{-1}$. However, dynamic strain ageing experiments should be extended for the different strain rate such as $5 \times 10^{-3} \text{ s}^{-1}$ and $10 \times 10^{-3} \text{ s}^{-1}$.

Transmission electron microscopy (TEM) can be used to resolve the clusters of C, N, Cr and Mo (2-3 nm). Further examination could be done to detect those clusters under favourable conditions using FEGTEM/EDX methods. Although hot tensile test results strongly suggest that the amount of interstitial carbon and nitrogen vary systematically in as-received samples, as-welded samples and weld metal depending on thermomechanical processing and welding conditions, a direct estimate of the contribution that interstitial solutes carbon and nitrogen make to the flow stress is lacking. It is also suggested that a further investigation of internal friction

measurement could be done using a torsional pendulum to get more information on diffusivities and precipitation of interstitial atoms such as C and N.

From as-received samples, as-welded samples and weld metal samples, impact test specimens could be obtained to study the changes in impact behaviour correlated with expected differences in dislocation and precipitate particles or solute atoms interactions. This experiment would give more informations about dynamic strain ageing, therefore, suggest a range of values for the cluster, solid solution and precipitate hardening contribution to flow stress of as-received samples, as-welded samples and weld metal.



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RESUME

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