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Microstructure And Mechanical Properites Of Welded Advanced Materials For Automotive Applications

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**MICROSTRUCTURE AND MECHANICAL
PROPERTIES OF WELDED ADVANCED MATERIALS
FOR AUTOMOTIVE APPLICATIONS**

By

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A thesis
presented to Ryerson University
in partial fulfillment of the
requirements for the degree of
Master of Applied Science
in the Program of
Mechanical Engineering

Toronto, Ontario, Canada, 2012

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AUTHOR'S DECLARATION

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ABSTRACT

Growing use of advanced materials (advanced high strength steel DP980 and ultralight-weight magnesium alloys) and innovative joining techniques (new-generation laser welding technology and weld bonding technique) is crucial for better fuel economy and lower CO₂ emissions in automotive manufacturing. Microstructures and mechanical properties of fiber laser welded high strength low alloy and DP980 steel joints, weld-bonded Mg/Mg and Mg/steel joints and adhesive-bonded Mg/Mg joints were studied. Tempered martensite and welding concavity were observed in fiber laser welded DP980 joints which reduced fatigue resistance, while both HSLA and DP980 joints showed a superior tensile strength. Weld-bonded Mg/Mg and Mg/steel joints with an adhesive layer were significantly stronger than resistance spot welded Mg/steel joints. Reducing bonding length on weld bonded Mg/Mg joints led to a higher maximum tensile shear stress, both tensile and fatigue strength were slight lower than that of adhesive bonded Mg/Mg joints, while ability of energy absorption was equivalent. The tensile properties reduced at a higher temperature (90°C) but it increased at a lower temperature (-40°C).

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To

My Parents Gaohai Xu and Qiuying Yu

Also my wife Wanna Li

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NOMENCLATURE

<u>Acronym</u>	<u>Definition</u>
3D	Three-dimensional
Al	Aluminium
AB	Adhesive-bonded (joint)
AC	Alternate current
AHSS	Advanced High Strength Steels
ASTM	American Society for Testing and Materials
BCT	Body-centered tetragonal structure
BM	Base metal
C	Carbon
CAFE	Corporate Average Fuel Economy
CCT	Continuous-cooling transformation (diagram)
CDZ	Columnar dendritic zone
Cr	Chromium
DC	Direct current
DLW	Diode laser welded (joint)
DP	Dual-Phase (steel)
EDS	Energy dispersive X-ray spectrometry
EDZ	Equiaxed dendritic zone
FEA	Finite Element Analysis
FLW	Fiber laser welded (joint)
FLW-M	Fiber laser welded multiple-linear joint
FLW-S	Fiber laser welded single-linear joint
FZ	Fusion zone
GI	Hot dip galvanizing coated (steel sheet)
G/R ratio	Ratio of temperature gradient (G) to growth rate (R)
HAZ	Heat Affected Zone
HSLA	High Strength Low Alloy

IMC	Intermetallic compound
LWB	Laser Welded Blank
Mg	Magnesium
MIG	Metal Inert Gas (welding)
Mn	Manganese
Mo	Molybdenum
N	Nitrogen
Nd: YAG	Neodymium-Doped Yttrium-Aluminum-Garnet
RD	Rolling direction
RM	Room temperature
RSW	Resistance spot welded (joint)
SEM	Scanning Electron Microscopy
Si	Silicon
S-N	Stress vs. Number of cycle to failure
TIG	Tungsten Inert Gas (welding)
TWBs	Tailor-Welded Blanks
UTS	Ultimate tensile strength
WB	Weld-bonded (joint)
XRD	X-ray diffraction analysis
YS	Yield strength
Zn	Zinc

<u>Symbol</u>	<u>Definition</u>
2θ	XRD diffraction angle, °
R	A load ratio of fatigue test, ($P_{\min}/P_{\max}$)
σ_a	Stress amplitude, MPa
σ_f'	Fatigue strength coefficient
N	Number of cycle to failure
b	Fatigue strength exponent
Q	Heat energy, J
I	Welding current, A
t	Welding time, s
R'	Dynamic resistance, ohm
$\dot{Q}$	Change of thermal energy Q with respect to time t , (dQ/dt),
h	Heat transfer coefficient
A	Surface area of the heat being transferred, m ²
T	Object's surface temperature, K
T_{env}	Temperature of the environment, K
$\bar{q}$	Local heat flux, W/(m ²)
k	Material's conductivity, W/(m·K)
∇T	Temperature gradient, K/m

CHAPTER 1 INTRODUCTION

Due to huge concerns on the global warming caused by recently recognized man-made greenhouse gas emissions and rising energy prices, demands for environmental-friendly transportation vehicles with better fuel efficiency and reduction of atmospheric pollution have received considerable attention [1-5]. To achieve such goals and meet legislative regulations, automotive manufacturers have to count on reducing the vehicle weight via applying advanced materials and innovative joining techniques [6-9].

Advanced materials for lightweighting applications are helping automakers manufacture lighter and more fuel-efficient vehicles. With the stronger steel, and lighter aluminum and magnesium components the current crop of cars and trucks can meet stringent crash safety standards, while also improve fuel economy to help achieve the more strict government regulations such as Corporate Average Fuel Economy (CAFE) mileage requirements [10].

In the past 10 years, the development and application of ultralight-weight magnesium alloys have been significantly increasing in the transportation sectors due to their low density, high strength-to-weight ratio, and superior damping capacity [11-16]. Another strategy that carmakers are using to produce lighter motor vehicles is the advanced high strength steel (AHSS) such as dual-phase (DP) steel for automotive body fabrication [17-24]. DP steels are a family of steels with a composite microstructure made of islands of martensite and sometimes bainite in the ferrite matrix. This type of microstructures provides the material with a higher initial work hardening rate and excellent uniform and total elongations to give DP steels a much higher ultimate tensile strength (UTS) and a lower yield strength (YS)-to-UTS ratio compared to the conventional steels with similar strengths. These characteristics have made DP steels attractive for automotive applications [25-27].

During manufacturing of automotive components, welding and joining are unavoidable. Laser welding is an enabling technology which is fast and precise in joining a wide variety of materials with varying thicknesses and types [28,29]. A study reported in 2008 showed that laser welding has partially replaced resistance spot welding for riveting of aircraft fuselage structures at

companies such as Airbus, and has been adopted at automobile manufacturers including Audi, BMW, and Volkswagen [30]. Thus, the research and development of new generation laser technology and its application such as fiber laser welding on DP steel i.e. DP980 is momentous and attractive to automotive manufacturing industry.

On the other hand, weld-bonding is an innovative and advanced hybrid joining technology since it has the advantages of both resistance spot welding and adhesive bonding, including reduced manufacturing costs, superior static and fatigue properties, improved corrosion resistance [31-36]. Indeed, it has been reported that automotive manufacturers such as Audi, Volkswagen, Peugeot Renault, Volvo, and BL have an active interest in bonding as a process to potentially replace resistance spot welding in their product since 1980's [37]. The first application of precision robot-dispensed two-part adhesives for bonding body-in-white structure was investigated by Mortimer [38] in 2004. Accordingly, a new attempt of weld bonded (WB) Mg-to-Mg and Mg-to-steel joints could bring benefits of motor vehicle weight reduction due to the combination of new WB joining technique with strategic ultralight-weight magnesium alloys. Therefore, to identify the influence of adhesive addition on the microstructures, tensile and fatigue properties of WB Mg/Mg and Mg/steel joints is highly technologically relevant to the automotive industry.

CHAPTER 2 LITERATURE REVIEW

2.1 Laser Welding Technology

Lasers have played an important role in the joining of materials since the invention of high power gas and solid state lasers in the 1960s [39-43]. Laser welding is used widely in the automotive manufacturing industry attributed to its ability to focus the intense beam (heat source) and reduce the weld bead width and the heat-affected zone (HAZ) regions [44]. Several types of laser welding are used most popularly today in the automotive, aerospace, shipbuilding, semiconductor, medical and jewellery industries, such as CO₂ laser welding, Nd:YAG laser welding, and diode laser welding [30].

The laser works by exciting in the atoms of the lasing medium (such as certain crystal, glasses, gases, and semiconductors, etc.) which must have a specific population inversion. When the excited atoms return to its ground state photons are emitted at a specific wavelength depending on the lasing medium. The photons that are reflected within the cavity cause more photons to produce, and part of the photons are released to form the laser beam which is focused by optics and delivered to the specific application [45-47].

The laser chamber or optical cavity contains an active lasing medium which usually determines the name of each laser. CO₂ laser uses a mixture of gases including CO₂ as the lasing medium and produces light with a 10.6 micron in wavelength; an artificial crystal is used for the lasing medium of Nd: YAG laser and 1.06 micrometer in light wavelength is produced. The lasers with high energy density such as CO₂, Nd: YAG and fiber are adoptable for keyhole welding. For many years, the CO₂ laser remained the only laser type of choice when deep penetration keyhole welding was required. A significant step forward was made when continuous wave solid-state lasers, in the form of the lamp pumped Nd:YAG laser, became commercially available, covering at least part of the power range of CO₂ lasers, and introducing the benefits and advantages of optical fiber delivery of the laser beam to the workpiece [48]. Although CO₂ and YAG lasers have dominated the laser-welding market, changes are underway because of the revolution and

development in high-conversion (power) efficiencies or high-power laser technology such as fiber laser technology [30].

A semi-conductor diode material is used as the active medium for diode lasers can be manufactured to produce one of several wavelengths. The laser is adoptable (or useful) for welding or laser brazing. It can have a maximum power of up to 6 kW, generate a near infrared light beam with a wavelength of 1.06 μm and is suitable for processing material without the use of beam transporting optics like the other lasers [45], in the case of diode lasers; the laser beam can be delivered to work pieces by optical fibers. Also, diode lasers can be mounted on coolers that do not interfere with the beam, which increases the efficiency to 10~20% whereas lamp pumped Nd:YAG laser lasers are typically 3-5% efficient [30,45,]. On the other hand, fiber lasers can easily achieve power levels of 5 kW up to 20 kW, with beam product as small as 1.8 mm mrad compared with YAG laser technology offering beam product of around 12 mm mrad [30]. It also has multiple advantages including flexible beam delivery, low maintenance costs, high efficiency, and compact size [43,49,50]. The comparison of above laser systems is shown in Table 2.1.

Table 2.1 Comparison of different laser systems [48].

Laser system	Laser medium	Average output power (kW)	Wavelength (μm)	Typical beam quality (mm mrad)	Efficiency	Welding modes
CO ₂ laser welding	Carbon dioxide gas mixed with helium and nitrogen	Up to 15	10.6	3.7	5~8%	Conduction and keyhole
Nd:YAG laser welding	Neodymium-doped yttrium aluminum garnet	Up to 6	1.06	12	3~5%	Conduction and keyhole
Diode laser welding	Semi-conductors	6	1.06	< 12	10~20%	Conduction
Fiber laser welding	Doped fiber	Up to 20	1.07	1.8	20~30%	Conduction and keyhole

2.2 Tailor-Welded Blanks

Although Honda had produced a tailored blank for a body side ring as early as in 1967, the tungsten inert gas (TIG) welding was not so well adopted for Tailor-Welded Blank (TWB). The development of laser welding made the TWBs technology “real start” in the 1980s, after which the laser-welded tailored blank was initially utilized [51]. TWBs were used in the automotive industry to consolidate parts, reduce tolerances, save weight and increases stiffness [52]. TWBs are steel sheets of different thickness and grades laser welded into a single flat blank prior to pressing to achieve the optimal materials arrangement and weight reduction for cars, and to increase process efficiency and machine flexibility [53]. Figure 2.1 shows typical applications of TWBs in a vehicle [54]. These TWBs consist of different grades and/or thicknesses of steel welded together to form optimized blanks that take advantage of the localized materials properties.

Fig. 2.1 Typical applications of TWBs in a vehicle [54].

According to the press release of a material supplier, about 15% of the body structure of a car is made of TWBs parts, which will increase to 25–30% in the next 5–10 years [55,56]. Recently,

Rooks [51] investigated the application of laser welded blank (LWB) for three-piece blanks (multiple linear welding) in the UK and possibility for four and five pieces LWB. While the integrity and durability of TWBs developed by fiber laser welding of the single linear joint and the multiple linear joint especially under cyclic loading are unknown, limited information was reported in the public literature.

2.3 HAZ Soft Zone Issue in the Laser Welding of DP Steels

Previous studies on DP steels showed that the welding led to the formation of a soft zone in the sub-critical area of the HAZ, and the mechanical properties and formability of the welded joints were significantly affected by the presence of this area [22,23,27,57-62]. Sreenivasan *et al.* [58] investigated the mechanical properties and performance of DP980 steel TWBs manufactured using Nd:YAG and diode laser welding. Formability and uniaxial tensile strength of DP980 welded blanks significantly reduced due to the occurrence of HAZ-softening, i.e., the hardness at outer HAZ being significantly below that of the base material caused by tempering of martensite. Similar effect of HAZ-softening on the formability was observed by Xia *et al.* [59] in their comparative study of high strength low alloy (HSLA) and DP980 steel joints prepared using diode laser welding. The study suggested a lower formability of DP980 diode laser welded (DLW) blanks compared to the base metal (BM), which could be improved by increasing welding speed which in turn reduced the HAZ-softening. In addition, Farabi *et al.* [22,58-60] have recently studied microstructure and mechanical properties of similar and dissimilar DLW joints in DP600 and DP980 and observed a significant change in the microstructure and hardness in the HAZ and fusion zone (FZ), which caused a reduction of the tensile and fatigue performance.

To reduce or minimize the effect of the soft zone on the mechanical properties, Fiber laser welding, which was first used in materials processing in 2000 [63], would be a potential solution for joining the stronger grades of automotive steels because its high power density, small beam divergence and focal spot diameter could lead to welds with a higher penetration and smaller width [43]. For example, results have been reported by Chowdhury *et al.* [64,65], where the width of HAZ and FZ after fiber laser welding was much smaller than that after diode laser

welding or double-sided arc welding in an AZ31B-H24 magnesium alloy. However, there is no report so far on the effect of fiber laser welding on the microstructure and mechanical properties of the important automotive steels namely HSLA and DP980. It is unclear how serious the HAZ softening in the HSLA and DP980 steels after fiber laser welding is, and if the multiple-linear fiber laser welding would have a significant effect on the fatigue strength.

2.4 Weld-bonding Process

Weld bonding which combines resistance spot welding with adhesive bonding was originally developed to prevent noise and vibrations for aircraft, automobile production and railway carriages [35]. WB joints were first used in USSR in planes of the type AN-24 [66]. The Soviet Union initially perfected this technology, which was known there as “glue welding”. The approach was a “flow-in” method as seen in Fig. 2.2(a), parts were welded together first, then the adhesive was flowed into the joint. A low-viscosity adhesive is used which penetrated the overlap joint by capillary action and is subsequently cured. The technique used in the United States is the “weld-through” method as seen in Fig. 2.2(b), the adhesive is applied to the parts first, and then spot welding is performed, finally the weldment is cured [36].

Since weld bonding has the advantages of both resistance spot welding and adhesive bonding, it has gathered wide acceptance as an effective joining process for significant improvement in static, dynamic and impact toughness properties of sheet metal joints. It also improves the corrosion, noise resistance and joint stiffness [67]. Some studies related to microstructures and mechanical properties of WB joints have been reported [31,32,68,69]. Chang *et al.* [31] studied a computational model and the stress field characteristics of WB joints for a car body steel sheet and observed a serious stress concentration existing at the edge of the weld spot in the spot-welded joints, and it could be removed or reduced by the application of adhesives. Goncalves & Marktins [32] examined static and fatigue performance of WB stainless steel joints which was used to assist project and design engineers in choosing to replace conventional spot-welded and/or adhesive bonded (AB) with WB joints. Zhang *et al.* [68] performed a comparison of mechanical properties and microstructure of weld nugget between WB and spot-welded DP steel, and concluded that weld bonding is a good option for joining sheets of DP780 steel. Weber *et al.*

[69] investigated resistance spot welding and weld bonding of AHSS and observed that the fatigue behavior has shown that the WB joints have an advantage over the RSW joints. As the process of WB is understood and its mechanical properties developed, the degree of acceptance of its applications has been increasing [36,68].

Fig. 2.2 Weld bonding: (a) the flow-in technique (1, weld; 2, apply adhesive, heat); (b) the weld-through technique (1, apply adhesive; 2, assemble; 3, spot weld, heat) [36].

2.5 Geometry and Temperature Effect in Weld-bonding

It has been noticed that the joint geometry is one of the most important issues for AB joints [70-73]. Goglio and Rossetto [70] studied an influence of geometry in the strength of AB steel joints and reported that the stress intensity decreases monotonically with increasing bonding length. da Silva *et al.* [71] studied the effect of geometry on the shear strength of AB DIN St33 steel (a German DIN standard steel) to DIN C65 steel (a German DIN standard steel) joints and observed the lap shear strength (i.e., in the form of failure load: kN) of the joint increased almost linearly

with overlap. Fessel *et al.* [72] evaluated different AB joint geometries for automotive applications using FEA and concluded that the stress distribution largely depended on the geometry of a joint. Pirondi and Moroni [73] investigated fatigue behavior of AB 6082-T4 aluminum and S255 steel joints and noticed that at the same level of average shear stress, the shorter the joint, the longer the duration. While there are some studies on the geometric influence of AB joints in the open literature, it is still not fully understood on how the joint geometry influences the microstructure and mechanical properties of WB joints.

On the other hand, it is given that the car body may experience temperature variations from -40°C to 90°C [74], it is necessary to ensure that the joints have a sufficient strength in that range of temperature. Many studies have been done in AB joints under a variety of environmental temperatures [74-76]. Grant *et al.* [74] studied the effect of temperature (-40°C , 20°C and 90°C) on the strength of AB single lap and “T” joints of steel sheet with a thickness of 0.95 mm for the automotive industry. Chen & Li [75] investigated the dynamic strength of adhesively bonded single lap joint of steel plate with a thickness of 1.7 mm at different temperatures (-40°C , 20°C and 90°C). Messler *et al.* [76] measured the effect of temperature on the static shear strength and fatigue strength for the laser beam weld bonding of Al alloy AA5754. However, it is still not clear how environmental temperature influences tensile properties of WB Mg/Mg, WB Mg/steel and AB Mg/Mg joints, since no tensile properties of such welds in relation to the temperature variation has been reported in the literature. Therefore, it is necessary to identify the influence of temperature on the tensile shear strength of WB joints.

2.6 Objectives

The objective of this thesis is to evaluate microstructures, tensile and fatigue properties of high power fiber laser welded (FLW) HSLA and DP980 steel joints which are compared with the DLW joints. Also, the WB Mg/Mg joints and Mg/steel joints are examined and compared with RSW Mg/steel joints and AB Mg/Mg joints, with special attention to the effect of various welding parameters on the microstructural change and mechanical properties.

Chapter 4 is aimed at evaluating the microstructural changes and fatigue resistance of FLW blanks with single (FLW-S) and multiple (FLW-M) welds of a DP980 steel, in comparison with those from the DLW blanks; Chapter 5 is aimed at evaluating and comparing the microstructure, hardness, tensile properties, and fatigue resistance of high power FLW HSLA and DP980 steel joints. The effect of weld geometry, namely single linear and multiple linear welding, on the microstructure and durability of the welds is also examined. Chapter 6 is to identify the influence of adhesive added on the microstructures, tensile and fatigue properties of WB Mg/Mg and Mg/steel joints, in comparison with RSW Mg/steel dissimilar joints. Chapter 7 is to evaluate the influence of joint geometry on the microstructures, monotonic and fatigue strength of WB Mg/Mg joints. Chapter 8 is to examine the influence of elevated (90°C) or low temperatures (-40°C) on the tensile shear strength in the WB Mg/Mg and Mg/steel joints with different joint geometries.

CHAPTER 3 EXPERIMENTAL PROCEDURE

3.1 Materials and Compositions

The starting materials in the present study for laser welding were hot-dip galvanizing (GI) coated HSLA and DP980 steel with a thickness of 1.2 mm, with the chemical compositions of the steels given in Table 3.1. The materials used in this study for WB, RSW and AB joints were GI coated HSLA steel sheet and Mg AZ31B-H24 alloy sheet, with a thickness of 2.0 mm and 0.7 mm, respectively. The chemical compositions for the HSLA steel and AZ31B Mg alloy are shown in Table 3.2 and 3.3, respectively. The geometry and dimensions of the metal sheets prepared for laser welding are shown in Fig. 3.1(a). The test coupons of 35 mm in width and 100 mm in length were cut for WB, RSW and AB joints.

Table 3.1 Compositions of HSLA steel and DP980 steel in the laser welding.

Steel grade	C	Mn	Si	Al	Mo	Cr	N
HSLA (wt.%)	0.0795	0.827	0.454	0.048	0.007	0.033	0.007
DP980 (wt.%)	0.15	1.5	0.31	0.05	-	-	-

Table 3.2 Compositions of HSLA steel in the weld-bonding.

Element	C	Si	Mn	P	S	Ni
Content (wt.%)	0.060	0.227	0.624	0.006	0.004	0.013
Element	Cr	Mo	Cu	Al	Ti	Nb
Content (wt.%)	0.041	0.005	0.044	0.039	0.003	0.021

Table 3.3 Compositions of Mg AZ31B-H24 alloy in the weld-bonding.

Element	Al	Zn	Mn	Si	Fe	Mg
Content (wt.%)	2.92	1.09	0.3	0.01	-	Balance

Fig. 3.1 Geometry and dimensions of work pieces and test specimens, (a) work pieces to be welded using a 6 kW fiber laser, (b) tensile and fatigue test samples sectioned from the single linear weld indicated by the short-dashed lines at (a), (c) tensile and fatigue test samples cut from the multiple linear weld indicated by the long-dashed lines at (a), and (d) a table of the specimen dimensions.

3.2 Welding Parameters

Laser welding was conducted using a Nuvonyx diode laser system and an IPG Photonics YLS-6000 fiber laser system, with the welding parameters used to fabricate the LWB shown in Table

3.4. The fiber laser system possessed a fiber core diameter of 0.3 mm and welding was performed with a head angle of 0° (i.e., the laser beam was perpendicular to the surface of work pieces); no shielding gas was used during the welding process.

Table 3.4 Details of the DLW and FLW parameters used for laser joining steels.

Laser system	Laser source	Laser power (kW)	Welding speed (m/min)	Focal length (cm)	Beam dimension (mm)
Nuvonyx ISL-4000	Diode	4	1	9	12×0.9
IPG Photonics YLS-6000	Fiber	6	16	30	0.6 (spot size in diameter)

Specimens for the WB, RSW and AB joints were ultrasonically cleaned for 5 minutes in acetone. Prior to welding, the surfaces of Mg coupons were further cleaned with a solution of 2.5% (weight/volume) chromic acid to remove surface oxides. For adhesive bonding, Terokal® 5087-02P adhesive was applied and then cured at a temperature of 180°C for 30 minutes. For WB process, resistance spot welding was performed immediately after adhesive applied on the workpieces. The welding parameters were selected to be 8 kA, 3 cycles plus 24 kA, 8 cycles using alternate current (AC) for WB Mg/Mg (*weld-bonded AZ31 to AZ31 Mg alloy*) joints; 8 kA, 3 cycles plus 24 kA, 8 cycles using direct current (DC) for WB Mg/steel (*weld-bonded AZ31 Mg alloy to HSLA steel*) joints; and 28 kA, 8 cycles with DC for RSW Mg/steel (*resistance spot welded AZ31 Mg alloy to HSLA steel*) joints. Symmetrical electrode cap (FF25) with a sphere radius of 50.8 mm and a face diameter of 16 mm was used for WB Mg/Mg similar welding. Asymmetrical electrodes, with a FF25 electrode cap against Mg side and a flat face electrode against steel side, were used for both WB Mg/steel and RSW Mg/steel dissimilar welding, aiming to balance the workpiece heating by reducing current density and increasing cooling rate of the steel sheet. For the study of geometry and temperature influence, different geometries

(bonding length or overlap areas) of the WB, RSW and AB joints are prepared and the details are shown in Table 3.5.

Table 3.5 Details of the geometry for the WB, RSW and AB joints.

Joint name in the thesis	Joint method	Overlap area (mm × mm)
WB-1 Mg/Mg joint	Weld-bonding	35 × 35
WB Mg/steel joint	Weld-bonding	35 × 35
RSW Mg/steel joint	Resistance spot welding	35 × 35
WB-0.5 Mg/Mg joint	Weld-bonding	35 × 17.5
AB-0.5 Mg/Mg joint	Adhesive-bonding	35 × 17.5

3.3 Metallographic and SEM Observations

The metallographic samples of laser welded joints for the microstructural examination were cut from the weld cross-section, then mounted, ground, polished, and etched with a 2% Nital solution; the metallographic samples of WB, AB and RSW joints were sectioned across the weld centre parallel to the loading direction, and then cold mounted, ground, polished and etched with a solution of 4.2 g picric acid, 10 ml acetic acid, 70 ml ethanol, and 10 ml water. The etched samples were observed using a light microscope along with a Clemex image analysis system and JEOL JSM-6380 scanning electron microscope (SEM), equipped with Oxford energy dispersive X-ray spectroscopy (EDS) and three-dimensional (3D) fractographic analysis capacity.

3.4 XRD Analysis

A multi-functional PANalytical X-ray diffractometer was used to identify the formation of potential intermetallic compounds after Mg/steel dissimilar welding from the fracture surface (Mg side) after tensile tests. X-ray diffraction (XRD) was performed using $\text{CuK}\alpha$ radiation at 45 kV and 40 mA. The diffraction angle (2θ) at which the X-rays hit the sample varied from 20° to 90° with a step size of 0.05° and 3s in each step.

3.5 Microhardness Test

Vickers microhardness was measured on the polished samples across the weld using a computerized microhardness tester at a load of 200 g and a dwell time of 15 s. All indentations were adequately spaced to prevent any potential effect of strain fields caused by adjacent indentations. To ensure the validity of test results, two calibration tests were done using a standard reference test block prior to the microhardness tests on the welded joints.

3.6 Tensile Test

Tensile test samples of laser welded joints following ASTM-E8/E8M standard [77] were sectioned from the welded blanks in the rolling direction (i.e., perpendicular to the welding direction), the examples of which are indicated by the dashed line in Figure 3.1(a) with the geometry and dimensions of the test coupons shown in Figs. 3.1(b) and (c). The test coupons were machined in such a way that the weld was positioned at the center of the gauge length for the single linear weld (Fig. 3.1(b)), or two welds were located within the gauge length symmetrically with respect to the middle of the sample (Fig. 3.1(c)). To identify the effect of welding and its geometry on the tensile and fatigue properties, coupons of the BM were also prepared and tested. Tensile tests of laser welded joints were conducted using a fully computerized United tensile testing machine at room temperature (RT) over a range of strain rates from 1×10^{-5} to $1 \times 10^{-2} \text{ s}^{-1}$. An extensometer with a gauge length of 50 mm and a strain limit of 20% was used to measure the strain during the tensile tests. Two samples were tested at each strain rate.

Tensile tests were carried out at RT using a fully computerized United tensile testing machine at different crosshead speeds of 0.1 mm/min, 1.0 mm/min and 10.0 mm/min for each group (WB Mg/Mg joints, WB Mg/steel joints, and RSW Mg/steel joints). Two samples were tested at each crosshead speed. For the purpose of measuring the influence of test temperature, an environmental chamber was used and the temperature was maintained to an accuracy of $\pm 2^\circ\text{C}$. Two samples were performed at each temperature of -40°C , 20°C and 90°C , respectively, with a

crosshead speed of 1.0 mm/min for each group (AB-0.5 Mg/Mg joints, WB-0.5 Mg/Mg joints, WB-1 Mg/Mg joints, and WB Mg/steel joints).

3.7 Fatigue Test

Fatigue tests were conducted in accordance with ASTM E466 [78] on a fully computerized Instron 8801 servo-hydraulic testing system under load control. Tension-tension cyclic loading at a stress ratio of $R=0.1$ and 0.2 was applied on laser welding joints and bonding joints respectively. A frequency of 50 Hz and sinusoidal waveform were used in all the tests. After fatigue test, the fracture surface morphology was examined using a JSM-6380LV SEM equipped with Oxford EDS system and 3D fractographic analysis capacity to identify fatigue crack initiation sites and fatigue crack propagation mechanisms.

CHAPTER 4 MICROSTRUCTURE AND FATIGUE PERFORMANCE OF SINGLE AND MULTIPLE LINEAR FIBER LASER WELDED DP980 DUAL-PHASE STEEL

4.1 Microstructure

The overall weld cross-section profile and microstructural changes across a DLW and a FLW DP980 joints are shown in Figs. 4.1 and 4.2, respectively. It is seen that the cross-section of the welds (Figs. 4.1(a) and 4.2(a)) exhibited a significant microstructural change in the HAZ and FZ for both welded joints. The width of HAZ for the DLW DP980 joint (on average ~2 mm) was much wider than that of the FLW DP980 joint (~0.4 mm). Similarly, the width of the FZ for the DLW DP980 joint was approximately 3 mm, which was about six-fold wider than that of the FLW DP980 joint (~0.5 mm on average), as seen in Fig. 4.1(a) and Fig. 4.2(a). Similar results have been reported by Chowdhury *et al.* [64,65,79], where the width of HAZ and FZ after fiber laser welding were much smaller than those after diode laser welding or double-sided arc welding in an AZ31B-H24 magnesium alloy. The reason for the narrower HAZ and FZ in the FLW was attributed to its considerably smaller laser beam size (0.28 mm^2 for the FLW versus 10.8 mm^2 for the DLW, Table 3.3) and the resultant high power density, which allowed it to weld at a much faster speed and at a lower heat input than the DLW [23].

More detailed microstructural changes could be seen in (b)-(d) of Figs. 4.1 and 4.2. It is seen from Fig. 4.1(b) and Fig. 4.2(b) that DP980 BM consisted of typical martensite in the ferrite matrix, where the volume fraction of martensite was determined by a direct image analysis tool in SEM to be approximately 0.56. Similar microstructural characteristics in the high-strength DP980 DP steel have also been observed with a volume fraction of martensite being 0.52, 0.50 and 0.55, respectively [22,27,80]. It should be noted that both images on the BM shown in Fig. 4.1(b) and Fig. 4.2(b) were aimed to indicate that the same material was used in both the fiber laser welding and diode laser welding. Microstructural examination of the HAZ of both DLW and FLW joints revealed the presence of tempered martensite in the ferrite matrix, as shown in Fig. 4.1(c) and Fig. 4.2(c), respectively, where some extent of decomposition of the originally existing martensite in the BM (Fig. 4.1(b) and Fig. 4.2(b)) could clearly be seen. Tempered

martensite and the associated HAZ-softened zone in the laser welded DP980 steel have been reported in the earlier studies [22,58,59,62,81]. The temperature experienced in the HAZ-softened region was close to or below the critical temperature (A_{c1} at which austenite begins to form during heating) during welding, leading to a tempering effect in which the martensite phase present in the BM decomposed [22,25,58,59,62,81-83]. The microstructure of tempered martensite typically consisted of very small and uniformly dispersed cementite (Fe_3C) particles embedded within a continuous ferrite matrix. As seen in Fig. 4.1(c) and Fig. 4.2(c) only partial tempering occurred in the HAZ after the FLW. It is of interest to observe that more martensite decomposition occurred in the HAZ of the DLW joint than in the HAZ of the FLW joint, as compared Fig. 4.1(c) with Fig. 4.2(c). This was due to the fact that a much longer welding time arising from a much lower welding speed (16 m/min in the DLW vs. 1 m/min in the FLW, as seen in Table 3.3), in conjunction with an overall more heat input as indicated by the considerably wider FZ and HAZ after DLW (Fig. 4.1(a)), resulted in a more complete martensite tempering/decomposition in the DLW process. Similarly, the Nd:YAG laser welds also exhibited less tempering than diode laser welds [23]. This was also true in the resistance spot welding where the cooling rate was much faster [83]. The final microstructures in the FZ also depended on the heat input and cooling rate, based upon a continuous-cooling transformation (CCT) diagram of low carbon steel, where the microstructure could contain a combination of grain boundary ferrite, side-plate ferrite, acicular ferrite, bainite and martensite as the FZ was cooled down to room temperature [84]. The SEM micrographs shown in Fig. 4.1(d) clearly showed such a microstructural feature consisting mainly of martensite, bainite and ferrite in the FZ of DLW. However, the FZ of the FLW joint exhibited almost a fully martensitic structure (Fig. 4.2(d)). While the FLW process had a higher power density than the diode laser welding process, the cooling rate in the FZ of the FLW joint was appreciably faster, due to the 16 times faster welding speed (Table 3.3) and much smaller FZ (Fig. 4.2(a)) in the fiber laser welding process. As discussed above, according to a CCT diagram of weld metal of low carbon steel [84], the thermal history path would move downward steeply at a rapid cooling rate missing the pearlite and bainite noses of the curve, so that the martensite start temperature would be directly reached. This means that there would be no chance to form other microstructure but possibly only martensite in the FZ of FLW joint, as shown in Fig. 4.2(d). Sreenivasan *et al.* [23] reported a

similar result for a 3 kW Nd:YAG laser welding applied on the DP980 steel sheets with a thickness of 1.17 mm.

Fig. 4.1 Micrographs showing the microstructure change of a DLW DP980 steel joint, (a) overall view of the cross section, (b) DP980 BM, (c) HAZ (soft zone), (d) FZ (where M: martensite, TM: tempered martensite, B: Bainite, and F: ferrite).

Fig. 4.2 Micrographs showing the microstructure change of a FLW DP980 steel joint, (a) overall view of the cross section, (b) DP980 BM, (c) HAZ (soft zone), (d) FZ (where M: martensite, TM: tempered martensite, and F: ferrite).

4.2 Hardness Profiles

Vickers microhardness profiles across the welds of the DLW joint and the FLW-S joint (as indicated by the dashed lines in Figs. 4.1(a) and 4.2(a)) are shown in Fig. 4.3. The hardness value in the DP980 BM was determined to be approximately 350-360 HV, which could be seen from the hardness profile of the FLW-S joint. After welding the hardness became obviously higher in the FZ of both DLW and FLW joints, and soft zones occurred beside the FZ in both types of laser welded joints (Fig. 4.3). However, the hardness in the FZ was much higher in the FLW joint (~480 HV) than in the DLW joint (~340 HV) due to the formation of predominantly martensitic structure in the FZ of FLW joint (Fig. 4.2(d)) with respect to the co-existence of

martensite, ferrite and bainite in the FZ of DLW joint (Fig. 4.1(d)). It is clear that the soft zone of FLW joint was substantially narrower than that of DLW joint. Besides, the former had a higher minimum hardness valley of about 280 HV than the latter (~200 HV). This was a consequence of a milder extent of tempering or decomposition of pre-existing martensite in the DP980 steel in the FLW joint (Fig. 4.2(c)) than in the DLW joint (Fig. 4.1(c)), stemming from a faster welding speed and cooling rate in the FLW process, since the tempered martensite had a lower hardness or strength than the normal martensite with a body-centered tetragonal (BCT) structure.

Fig. 4.3 Hardness profiles across a FLW DP980-S and a DLW DP980 joint.

Fig. 4.4 shows the microhardness profile across the weld of a FLW multiple linear joint. Each weld bead in Fig. 4.4 had a similar result to that of the FLW single linear joint (Fig. 4.3), i.e., with a FZ hardness of ~480 HV and an average soft zone hardness of ~280 HV. The only difference lay in that the multiple linear FLW joint had four soft zones located at both sides of two FZ pylons. It would be of interest to see if these multiple soft zones would affect tensile and fatigue behavior of the FLW joint.

Fig. 4.4 Hardness profile across a FLW DP980-M joint.

4.3 Tensile Properties

Fig. 4.5 demonstrates the typical engineering stress versus engineering strain plots of the BM, DP980, DLW DP980 joint, FLW DP980-S joint, and FLW DP980-M joint tested at a strain rate of $1 \times 10^{-3} \text{ s}^{-1}$. It is seen that while the strain to failure was lower, the YS after FLW remained very close to that of BM DP 980 steel. However, the YS after DLW decreased remarkably. The evaluated tensile test results are listed in Table 4.1. The YS and UTS of the DLW DP980 joint was obtained to be an average of 505 MPa and 724 MPa, respectively, which were much lower than those of the BM having a YS of 720 MPa and a UTS of 1095 MPa. Interestingly, the tensile properties of both FLW DP980-S joint (YS = 719 MPa; UTS = 1048 MPa) and FLW DP980-M joint (YS = 717 MPa, UTS = 1066 MPa) were very close to the values of the BM, and no obvious difference of the tensile properties between the FLW DP980-S and DP980-M joints was observed. The joint efficiency reached as high as about 96~97% (Table 4.1), where the joint efficiency is normally defined as a ratio of the UTS of welded joints to the UTS of the

corresponding BM [64,65,79]. Indeed, if one considers a ratio of the YS of welded joints to the YS of the corresponding BM, then the ratio would arrive at nearly 100%. Thus, it could be concluded that the FLW joint had a significantly higher tensile performance than the DLW and was nearer almost to that of the BM. It was ascribed to the higher welding speed (Table 3.3) and lower heat input in the fiber laser welding which greatly reduced the detrimental influence of the soft zone by narrowing it down significantly compared to the diode laser welding which resulted in a wider soft zone (Fig. 4.1(a) and Fig. 4.3). The narrower HAZ was better restrained by the neighbouring material which restricted thinning [85].

Fig. 4.5 Representative engineering stress versus engineering strain of the DP980 BM, DLW DP980 joint, FLW DP980-S joint, and FLW DP980-M joint tested at a strain rate of $1 \times 10^{-3} \text{ s}^{-1}$.

Table 4.1 Tensile properties, joint efficiency, and fatigue parameters σ_f' and b for the DP980 BM, DLW DP980 joints, FLW DP980-S joints, and FLW DP980-M joints.

Welding type	Yield strength (MPa)	Ultimate tensile strength (MPa)	Elongation (%)	Joint efficiency (%)	Fatigue limit (MPa)	Fatigue ratio	σ_f' (MPa)	b
BM DP980	720	1095	14.2	-	250	0.228	1019	-0.098
DLW DP980	505	724	7.6	66.1	150	0.207	456	-0.054
FLW DP980-S	719	1048	4.6	95.7	150	0.143	1169	-0.132
FLW DP980-M	717	1066	5.1	97.3	<150	0.141	1984	-0.185

On the other hand, ductility of both the DLW and the FLW joints were lower than that of the BM. This was due to the fact that yielding occurred first in the soft zones and the subsequent plastic deformation was predominantly concentrated there. This led in turn to a reduction in the overall specimen elongation, as necking and failure were observed to occur in the soft zone. Similar results were reported in a study of formability of diode laser welding DP980 and HSLA steels [62,81]. This would also explain the higher ductility in the diode laser welding where the wider soft zone (Fig. 4.1(a) and Fig. 4.3) could accommodate larger strain before failure compared to the narrower soft zone in FLW (Fig. 4.2(a) and Fig. 4.3). It was also due to premature failure induced by the soft zone (i.e., lower ductility) that the UTS of FLW joints

became lower (Fig. 4.5), although a high joint efficiency of about 96~97% was achieved (Table 4.1).

4.4 Fatigue Strength and Failure Mode

Fatigue test results of the BM DP980, DLW DP980 joints, FLW DP980-S joints and FLW DP980-M joints obtained at $R=0.1$, 50 Hz, and RT are plotted in Fig. 4.6. The DLW DP980 joints were found to have a much lower fatigue life or fatigue strength than the BM at both high and low levels of stress amplitudes. For example, the fatigue strength at 1×10^7 cycles (sometimes called conditional fatigue limit) was about 100 MPa lower in the DLW DP980 joints than the BM, while it was further lower (~ 150 MPa) in the low cycle fatigue region at 2×10^3 cycles. This corresponded well to the effect of the wider soft zone with significantly lower values in both hardness (Fig. 4.3) and tensile strength (Fig. 4.5 and Table 4.1). The FLW-DP980 joints exhibited an equivalent or only slightly lower fatigue life than the BM at a stress amplitude above 300 MPa, whereas the fatigue strength of FLW joints became lower and more scattered below a stress amplitude of 300 MPa. This indicated that while the narrower HAZ-softening (Fig. 4.2(a), Fig. 4.3 and Fig. 4.4) did not affect the static tensile properties (Fig. 4.5 and Table 4.1), the dynamic fatigue resistance was still susceptible to the presence of soft zone, irrespective of the size of soft zones. This observation corresponded well to that reported by Roesler *et al.* [86] who noted that the fatigue strength of a material under dynamic cyclic loading was much more sensitive to the manufacturing process and materials than the static strength, and that the fatigue strength was also much more sensitive to the lower level cyclic load than the higher level cyclic load. Indeed, in the region of lower level of stress amplitude and longer life, the factors like surface conditions, residual stresses, localized stress concentration, surface protective coating, severe weld concavity and zinc inclusions in weldment are detrimental to fatigue life [61,87]. The fatigue data plotted in Fig. 4.6 for the multiple linear welds exhibited a larger scatter and lower fatigue strength (e.g., two samples failed at 150 MPa), indicating that the probability of dynamic fatigue failure at the lower level of stress amplitudes increased with increasing number of soft zones.

Fig. 4.6 S-N curves of the DP980 BM, DLW DP980 joints, FLW DP980-S joints, and FLW DP980-M joints tested at $R=0.1$, 50 Hz, and RT.

The fatigue limit and fatigue ratio are tabulated in Table 4.1. The fatigue limit of both the DLW DP980 joints and FLW DP980 joints were obtained to be 150 MPa, which was 40% lower than that of the BM, and the fatigue ratio (i.e., a ratio of fatigue limit to the UTS) of the DLW DP980 joints and the FLW DP980 joints were 0.21 and 0.14, respectively. The lower fatigue ratio of FLW DP980 joints was related to the higher UTS (Table 4.1). These results suggested that while the tensile strength of the FLW joints greatly improved, the fatigue strength only improved at a higher level of stress amplitudes (>300 MPa), and was equivalent to that of the DLW joints at a lower level of stress amplitudes (<300 MPa). One more factor for the low fatigue limit or fatigue ratio in both types of FLW DP980 joints was associated with the notch effect of weld concavity, since fatigue is a type of notch-sensitive dynamic failure. As seen in Fig. 4.2(a), the FLW “notch” or weld concavity was similar to a notched specimen (e.g., in [88]) where the stress concentration was generated. As a consequence, this would result in a decreased fatigue limit or

fatigue life at the low level of stress amplitudes, stemming from the presence of fatigue notch sensitivity.

The obtained fatigue data shown in Fig. 4.6 could be further fitted using the following Basquin type equation,

$$\sigma_a = \sigma_f' (2N)^b \quad (4.1)$$

where σ_a is the stress amplitude, σ_f' is the fatigue strength coefficient defined by the stress intercept at $2N=1$, N is the number of cycle to failure, and b is the fatigue strength exponent. The obtained values of σ_f' and b of the BM, the DLW DP980 joints and the FLW DP980 joints tested at $R=0.1$, 50 Hz and RT are given in Table 4.1, where the fatigue life at a given stress amplitude was dependent on both fatigue strength coefficient σ_f' and fatigue strength exponent b .

4.5 Fractography

The fatigue failure locations for the FLW DP980-S and FLW DP980-M joints are shown in Fig 4.7. Fatigue failure of all test samples was observed to occur predominantly at the HAZ soft zones. It was less likely to occur at the center of FZ due to the strong FZ hardness “pylon” effect in the FLW (Figs 4.3 and 4.4). The failure location of the FLW joints was similar to that of the DLW DP980 joints in which all the fatigue failure also occurred at the soft zones as reported in our earlier study [21].

Fig 4.7 Typical fatigue failure locations of (a) a FLW DP980-S joint, and (b) a FLW DP980-M joint.

SEM images of fatigue fracture surface of a FLW DP980-S joint tested at an applied stress amplitude of 350 MPa are shown in Fig. 4.8. It is seen from Fig 4.8(a) that multiple crack initiation occurred from the surface where the weld concavity met with the soft zone. The magnified SEM images on some features near a crack initiation site are shown in Fig. 4.8(b) and (c), where some gas pores trapped during FLW (Fig. 4.8(b)) and surface welding defects (Fig. 4.8(c)) could be seen, as indicated by arrows on the images. These were similar to fatigue fracture surface features reported for a FLW AZ31B-H24 Mg alloy [79]. The fatigue crack initiation in the present FLW DP980 steel would mainly be related to the severe weld concavity (Fig. 4.2(a)) and the ensuing stress concentration, and the presence of soft zones (Fig. 4.3 and Fig. 4.4). Fatigue crack propagation was mainly characterized by fatigue striations in conjunction with secondary cracks, as shown in Fig. 4.8(d), while the final fast propagation area consisted of characteristic dimples. The spacing of fatigue striations increased with increasing distance from the crack initiation site, reflecting a gradually increasing fatigue crack propagation rate. Fatigue striations basically occurred via a repeated plastic blunting-sharpening process via the slip of dislocations in the plastic zone ahead of the fatigue crack tip [89].

Fig. 4.8 Typical SEM images of fatigue fracture surface of a FLW DP980-S joint tested at a stress amplitude of 350 MPa, (a) overall view, (b) fracture surface at a larger magnification, (c) crack initiation area, (d) crack propagation area, (e) final fast propagation area.

4.6 Summary

- (1) Both DLW and FLW joints showed tempered martensite at the outer HAZ which caused softening in the welds. More complete martensite tempering or decomposition occurred in the HAZ of the DLW joint than in the HAZ of the FLW joint due to a lower welding speed and more heat input.

- (2) While the microstructure in the FZ of DLW joints consisted mainly of martensite, ferrite, and bainite, almost fully martensitic structure was observed in the FZ of FLW joints due to a faster cooling rate. This led to a characteristic “suspension bridge”-like hardness profile with the FZ hardness as a “pylon” in the FLW.
- (3) The FLW joints had a much narrower soft zone with a higher hardness due to a higher welding speed and lower degree of martensite tempering, in comparison with the DLW joints. This resulted in a higher ultimate tensile strength with a joint efficiency reaching as high as 96~97% while the yield strength was essentially unaffected and ductility became lower.
- (4) At a higher stress amplitude above 300 MPa, fatigue life of the FLW joints was equivalent to that of BM, while fiber laser welding did not result in an improved fatigue performance at a lower stress amplitude below 300 MPa in comparison with the DLW joints, because the dynamic fatigue resistance was susceptible to the presence of soft zone and the weld concavity-induced fatigue notch sensitivity. The multiple linear welds appeared to increase the probability of dynamic fatigue failure at a lower level of stress amplitudes due to an increasing number of weld concavities and soft zones present.
- (5) Fatigue crack initiation was observed to occur from the specimen surface where the weld concavity met with a soft zone, and crack propagation was characterized by the characteristic fatigue striations coupled with secondary cracks for all types of the welded joints.

CHAPTER 5 TENSILE AND FATIGUE PROPERTIES OF FIBER LASER WELDED HIGH STRENGTH LOW ALLOY AND DP980 DUAL-PHASE STEEL JOINTS

5.1 Microstructure Evolution

The representative weld cross-section profiles and microstructure at different weld zones of the HSLA and DP980 joints made with the FLW are shown in Fig. 5.1. The cross-section of the welds showed a significant microstructural change in both HSLA and DP980 joints, indicating the formation of the middle FZ and two adjacent HAZs which coexisted with the unaffected BM at both left and right ends as indicated in Fig. 5.1(a) and Fig. 5.1(b). It is seen that the width of HAZ and FZ in both HSLA and DP980 joints was similar with an average of 200~300 μm , and 400~500 μm , respectively. The HAZ and FZ were narrower when compared to those formed in other laser welding (Table 5.1). For instance, a 4 kW diode laser welding of 1.2 mm thick DP980 steel sheets resulted in 4 mm wide HAZ and 3 mm wide FZ [62]. This was attributed to considerably larger laser beam size (6 mm^2 for the diode laser welding versus 0.28 mm^2 for the fiber laser welding), where the lower energy density (less than 10^6 W/cm^2) produced conduction mode. In contrast, in the FLW process the higher energy density which caused keyhole mode made welding more efficient compared with diode laser welding [23,47,90], and the fiber laser welding could be operated at higher welding speed with a narrower weld (Table 5.1).

HSLA BM consisted of fine grained ferrite matrix with a uniform dispersion of fine alloy carbides (Fig. 5.1(c)), whereas martensite phase embedded in a continuous ferrite matrix was seen in DP980 BM (Fig. 5.1(d)). The volume fraction of martensite in the DP980 BM was estimated to be ~56% using image analysis in SEM, which matched closely with the earlier studies on the DP980 steel of similar chemistry [22,27,80,83]. Figs. 5.1(e) and (f) showed the microstructural changes of HAZ in both HSLA and DP980 joints, where the typical micrographs were taken from the HAZ positioned at the left side of FZ corresponding to “e” in Fig. 5.1(a) and “f” in Fig. 5.1(b). The details of HAZ microstructure will be discussed in the later section.

Fig. 5.1 Micrographs showing the microstructure change of a FLW HSLA and a FLW DP980 steel joint, (a) overall view of the cross section of the FLW HSLA steel joint, (b) overall view of the cross section of the FLW DP980 steel joint, (c) HSLA BM, (d) DP980 BM, (e) HAZ of the FLW HSLA steel joint, (f) HAZ of the FLW DP980 steel joint, (g) FZ of the FLW HSLA steel joint, (h) FZ of the FLW DP980 steel joint.

The microstructure in the FZ depended heavily on the heat input and cooling rate as discussed in Chapter 4. However, Figs. 5.1(g) and 5.1(h) clearly showed that fully martensitic structure formed in the FZ of both HSLA and DP980 joints after the present fiber laser welding. The reason was mentioned in Chapter 4 for this. It should be noted that similar microstructural change was observed in the case of the multiple linear welds except the presence of twice microstructural heterogeneities, i.e., FZ and HAZ, because of doubling the linear welds. Thus, their microstructure was not presented here.

Table 5.1 Comparison of the average width of HAZ and FZ in the welded joints made with different types of laser welding.

Welding type	Power (kW)	Welding speed (m/min)	Spot size (mm)	Welding materials	Thickness of workpiece (mm)	Average width of HAZ (μm)	Average width of FZ (μm)	Reference
DLW	4	1.6	12*0.5	DP980 steel	1.2	4000	3000	[64]
Nd:YAG Laser welding	3	3	0.6	DP980 steel	1.17	1000	750	[23]
CO ₂ laser welding	6	6	-	DP780 steel	1.8	1000	1000	[99]
FLW	6	16	0.6	DP980 steel	1.2	250	450	Present study

SEM images of the inter-critical HAZ (experiencing temperatures between A_{c1} and A_{c3} lines) with adjacent BM and upper-critical HAZ (region above A_{c3} line) of HSLA joint are shown in Fig. 5.2. The HSLA BM microstructure (Fig. 5.2(b)) was confirmed to consist of ferrite grains with finely dispersed alloy carbide particles as observed in optical image (Fig. 5.1(c)). Fig. 5.2(c) represented the microstructure containing some martensite which formed as a product of fast

cooling from the inter-critical region, where the partially formed austenite (i.e., co-existence of ferrite and austenite) transformed to martensite during fast cooling down to room temperature in the fiber laser welding process. However, it should be pointed out that the HAZ of HSLA joint did not involve any microstructural changes at the sub-critical region (below A_{c1} line) due to the absence of martensite in its BM.

Fig. 5.2 SEM micrographs showing the HAZ microstructure change in the FLW of HSLA steel, (a) overall view of inter-critical HAZ with adjacent BM and upper-critical HAZ, (b) HSLA BM with ferrite and fine alloy carbides, and (c) inter-critical HAZ showing newly formed martensite phase (M: martensite, and F: ferrite).

On the other hand, for DP980 steel after fiber laser welding a tempered martensite area (or sub-critical region below A_{c1} line) located in-between the BM and inter-critical zone (between A_{c1} and A_{c3} lines), along with an upper-critical HAZ (above A_{c3} line) could be seen in Fig. 5.3(a). While the microstructure in the inter-critical zone was somewhat similar to that in the BM with

ferrite plus martensite islands, in the tempered martensite area some extent of decomposition of the martensite pre-existed in the BM (Fig. 5.3(b)) was clearly visible at a higher magnification as shown in Fig. 5.3(c). The microstructure of tempered martensite in DP980 welds has been reported to consist typically of uniformly distributed fine precipitated cementite (Fe_3C) particles embedded within a continuous ferrite matrix [83,91-93]. As seen from Fig. 5.3(c), it was clear that only partial tempering of martensite occurred in the HAZ of DP980 in the fiber laser welding process.

Fig. 5.3 SEM micrographs showing the microstructural change in a DP980 steel joint, (a) overall view of tempered martensite area in the HAZ with adjacent BM, inter-critical HAZ, and upper-critical HAZ, (b) DP980 BM showing ferrite and martensite, and (c) tempered martensite region in the HAZ (M: martensite, TM: tempered martensite, and F: ferrite).

5.2 Hardness Profiles

Vickers microhardness profiles across the welds of the FLW HSLA-S and FLW DP980-S (as indicated by the red dashed lines in Fig 5.1(a) and (b)) are showed in Fig. 5.4. The average hardness value of the HSLA and DP980 steel BM was measured to be about 170 HV and 340 HV, respectively. The higher hardness in the DP980 BM was attributed to the presence of higher fraction of hard martensite phase (~56%, Fig. 5.1(d) and Fig. 5.3(b)), while the HSLA BM did not contain any martensite (Fig. 5.1(c) and Fig. 5.2(b)). In accordance with the microstructure (Figs. 5.1(g) and (h)) where fully martensitic structure emerged, the hardness of the FZ became much higher and showed a decrease from FZ to the HAZ. Interestingly, the hardness curve though the whole HAZ was seen to merge smoothly into the unaffected BM in the HSLA joint due to a decrease in the fraction of martensite in the HAZ which formed as a solid-state phase transformation from austenite in the inter-critical region.

On the other hand, the hardness pattern in the DP980 joint exhibited two soft zone “valleys” beside the center FZ, where the hardness locally dropped significantly to an average of 280 HV which was well below the DP980 BM hardness (~350 HV). This was attributed to the occurrence of martensite tempering (HAZ-softened region) as discussed above and also illustrated in Fig. 5.3. The narrow FZ with fully martensitic microstructure in both HSLA and DP980 joints after FLW were characterized by a very high hardness (Fig. 5.4). However, the DP980 joint had a further higher hardness averaging 450 HV compared with the average FZ hardness (360 HV) in the HSLA joint. This was mainly due to the fact that the martensite in the FZ of DP980 joint contained a higher amount of carbon, as seen in Table 3.1 where nearly twice carbon content was present in the DP980 steel compared with the HSLA steel. Additionally, nearly doubled manganese content in the DP980 steel (Table 3.1) also contributed to the higher hardness [94]. Calcagnotto *et al.* [95] noted that Mn was able to increase hardneability substantially just like carbon content in austenite on grain size stability and hardenability in ultrafine-grained ferrite /martensite DP steel.

Fig. 5.4 Representative hardness profiles across the HSLA-S and DP980-S joints.

Fig. 5.5 shows the microhardness profiles across the multiple linear weld of FLW HSLA-M and FLW DP980-M joints. Both FLW HSLA-M and FLW DP980-M joints were observed to display a “suspension bridge”-like or “cable-stayed bridge”-like hardness profile with two FZ hardness representing the “pylons or towers”. It is seen that each weld bead of the multiple linear joints (Fig. 5.5) had a similar hardness values when compared it to the single linear joints (Fig. 5.4), which was a result of similar microstructures as noted in the previous section. The distinction between the FLW HSLA-M and FLW DP980-M joints was that the hardness profile of the former was smoother without “valleys” due to the absence of HAZ-softening (martensite tempering). On the contrary, in the FLW DP980-M joint, each FZ hardness “pylon” was accompanied by two soft zone “valleys” positioned at its two sides (Fig. 5.5). It would be of interest to see if doubling the number of soft zones would affect tensile strength and fatigue life of the HSLA and DP980 joints.

Fig. 5.5 Typical hardness profiles across the HSLA-M and DP980-M joints.

5.3 Tensile Properties

The representative engineering stress versus engineering strain curves, determined at a strain rate of $1 \times 10^{-3} \text{ s}^{-1}$, of the BM, FLW single and multiple linear joints of both HSLA and DP980 steels are illustrated in Fig. 5.6. In the HSLA joints, both the strain to failure and the YS after welding were observed to be very close to those of the BM. The evaluated tensile test results are listed in Table 5.2. The YS and UTS were obtained to be 400 MPa and 515 MPa, respectively, for the FLW HSLA-S joint and 398 MPa and 523 MPa, respectively, for the FLW HSLA-M joint. These values were fairly close to those of the BM which had a YS of 455 MPa and UTS of 546 MPa (Table 5.2, and Fig. 5.6). These tensile results suggested that the present FLW process, regardless of single or multiple linear welding had little influence on the tensile properties of the HSLA steel. Obviously, this was because the FLW of HSLA steel did not create any soft zone (Figs 5.4 and 5.5) in the HAZ that was detrimental to the mechanical properties of the welds. This was also corroborated by the tensile failure of FLW HSLA joints that consistently occurred

in the BM. This finding was also in agreement with the earlier observations on the tensile failure locations by Xia *et al.* in a DLW HSLA joint [62], Farabi *et al.* [60] in a DLWDP600 steel joint, and Panda *et al.* [85] in a DLW HSLA/DP980 dissimilar joint with a thickness of 1.14 mm and 1.17 mm, respectively.

Fig. 5.6 Representative engineering stress versus engineering strain curves of HSLA BM, FLW HSLA-S joint, FLW HSLA-M joint, DP980 BM, FLW DP980-S joint, FLW DP980-M joint tested at a strain rate of $1 \times 10^{-3} \text{ s}^{-1}$.

In contrast to the HSLA welds, the DP980 joints showed a lower strain to failure (Fig. 5.6 and Table 5.2). However, the YS and UTS were basically unaffected by the FLW. As seen in Table 5.2, the strength of FLW DP980-S joints (YS = 719 MPa; UTS = 1048 MPa) and FLW DP980-M joints (YS = 717 MPa, UTS = 1066 MPa) was very close to those of the BM (YS = 720 MPa; UTS = 1095 MPa). The results indicated that the presence of narrow soft zone in the FLW DP980 welds was not detrimental to the tensile strength, and the high welding speed could greatly reduce the detrimental influence of the soft zone by narrowing it down significantly.

It was also of interest to observe that a joint efficiency (i.e., a ratio of the UTS of welded joints to the UTS of the corresponding BM [66,67,81]) of both FLW HSLA joints and FLWDP980 joints reached as high as about 94~96% and 96~97% (Table 5.2), respectively. These results were similar to those of Nd:YAG laser welding (94.1% for DP980 steel joints) [23] and CO₂ laser welding (96.9% for CP1180 steel joints [96]). Indeed, if one considers the ratio of the YS of welded joints to the YS of the corresponding BM, a nearly 100% YS joint efficiency would be achieved in both FLW DP980-S and FLW DP980-M joints. From these tensile results it can be concluded that FLW is an able and viable joining process that gave rise to superior tensile performance approaching that of BMs for both HSLA and DP980 steels, while the FLW DP980 joints had a much higher tensile strength than the FLW HSLA joints.

Table 5.2 Tensile properties and fatigue parameters σ_f' and b for HSLA BM, FLW HSLA-S joint, FLW HSLA-M joint, DP980 BM, FLW DP980-S joint, and FLW DP980-M joint.

Welding type	Yield strength (MPa)	Ultimate tensile strength (MPa)	Elongation (%)	Joint efficiency (%)	Fatigue limit (MPa)	Fatigue ratio	σ_f' (MPa)	b
HSLA BM	455	546	24.7	---	200	0.366	302	-0.023
FLW HSLA-S	400	515	22.7	94.3	125	0.243	328	-0.041
FLW HSLA-M	398	523	20.7	95.8	125	0.239	343	-0.048
DP980 BM	720	1095	14.2	---	250	0.228	1019	-0.098
FLW DP980-S	719	1048	4.6	95.7	150	0.143	1169	-0.132
FLW DP980-M	717	1066	5.1	97.3	150	0.141	1984	-0.185

However, it should be noted that, while the percent elongation or ductility of both FLW HSLA-S and HSLA-M joints (~21-23%) came fairly close to that of the HSLA BM (~25%) as shown in Table 5.2, the ductility of the FLW DP980-S joints (4.6%) and the FLW DP980-M joints (5.1%) was noticeably lower than that of the BM (14.2%) since all the tensile test samples failed consistently in the soft zone. The lower ductility in the DP980 welds was due to the fact that yielding occurred first in one of the soft zones and the subsequent plastic deformation was predominantly concentrated there leading to necking and premature failure at that location, which resulted in a reduction in the overall specimen elongation. Similar results have also been reported in the study of formability of DLW of DP980 [62,85]. It was also due to the premature failure induced by the soft zone (i.e., lower ductility) that the UTS of FLW joints became lower (Fig. 5.6), although a high joint efficiency of about 96~97% was achieved in the FLW DP980 joints.

5.4 Fatigue Strength and Failure Mode

Fatigue test results of the BM, FLW single linear and multiple linear joints of HSLA tested at $R=0.1$, 50 Hz, and RT are plotted in the Fig. 5.7(a). Both FLW HSLA-S and HSLA-M joints exhibited an equivalent fatigue life within the experimental scatter. At a stress amplitude of above 250 MPa the fatigue life of the joints was only slightly lower than that of the BM, whereas the fatigue strength became lower and more scattered at lower stress amplitudes (<250 MPa). In FLW HSLA welds, in spite of the absence of HAZ-softening (Figs 5.4 and 5.5), welding could destroy the surface continuance, change the surface residual stress distribution, and create the weld concavity on the surface (Fig. 5.1(a)). These could promote the fatigue crack initiation and propagation, thus reducing fatigue life [61,65,]. Interestingly, the multiple linear welded HSLA samples exhibited equivalent fatigue strength at both higher and lower levels of stress amplitudes to that of the single linear welds, indicating that the probability of dynamic fatigue failure was not affected by the weld geometry or by increasing number of welds, i.e., HAZ and FZ in the HSLA.

The S-N curves of DP980 BM, FLW DP980-S joints, and FLW DP980-M joints tested at $R=0.1$, 50 Hz, and RT are shown in the Fig. 5.7(b). The FLW DP980-S joints represented an equivalent

or only slightly lower fatigue life than the BM at stress amplitude above 250 MPa, whereas the fatigue strength became lower and more scattered below stress amplitude of 250 MPa. This indicated that while the narrow HAZ-softening (Fig. 5.1(b), Fig. 5.3(a), and Fig. 5.5) did not affect the static tensile strength (Fig. 5.6 and Table 5.1), the dynamic fatigue resistance was susceptible to the presence of the soft zone. This observation corresponded well to that reported by Roesler *et al.* [86] who pointed out that the fatigue strength of a material under dynamic cyclic loading was much more sensitive to the manufacturing process and materials than the static strength, and that the fatigue strength was also much more sensitive to the lower level cyclic load than the higher level cyclic load. Indeed, in the region of lower level of stress amplitude and longer life, the factors like surface conditions, residual stresses, localized stress concentration, surface protective coating, severe weld concavity and zinc inclusions in weldment are deleterious to fatigue life [61,79]. The fatigue data plotted in Fig. 5.7(b) for the multiple linear welds exhibited a larger scatter and lower fatigue strength (e.g., two samples failed at a stress amplitude of 150 MPa), implying that the probability of dynamic fatigue failure at the lower level of stress amplitudes increased with increasing number of soft zones in the higher grade of DP980.

To make a better comparison, fatigue test results of the FLW HSLA-S joints, the FLW HSLA-M joints, the FLW DP980-S joints, and the FLW DP980-M joints were plotted in Fig. 5.7(c). The FLW DP980 joints were observed to have a much higher fatigue strength than that of the FLW HSLA joints at a stress amplitude above 250 MPa, but the fatigue strength was equivalent below a stress amplitude of 250 MPa for both DP980 and HSLA joints due to the fact that the higher grade/strength of DP980 was more sensitive to the presence of severe weld concavity (Fig. 5.1(b)) especially at the lower level of cyclic load than at the higher level of cyclic load.

(a)

(b)

Fig. 5.7 S-N curves of (a) the HSLA BM, FLW HSLA-S joints and FLW HSLA-M joints, (b) the DP980 BM, FLW DP980-S joints and FLW DP980-M joints, and (c) FLW HSLA-S joints, FLW HSLA-M joints, FLW DP980-S joints and FLW DP980-M joint tested at $R=0.1$, 50 Hz, and RT.

Fatigue limit and fatigue ratio are tabulated in Table 5.2. The decrease in fatigue limit for both FLW HSLA and FLW DP980 joints, with respect to their BM, were similar at 37.5% and 40%, respectively, while the fatigue limit of the FLW DP980 joints were 20% higher than that of the FLW HSLA joints. The fatigue ratio (i.e., a ratio of fatigue limit to the UTS) of the FLW HSLA-S joints and the FLW DP980-S joints were 0.243 and 0.143, respectively. The lower fatigue ratio of FLW DP980-S joints was related to its much higher (or twice) UTS (Table 5.2). These results suggested that while the tensile strength of the FLW DP980 joints was much higher than that of the FLW HSLA joints, the fatigue strength was very close for both joints at a lower level of stress amplitudes (<250 MPa). Another factor for the lower fatigue limit or fatigue ratio in both types of FLW joints was associated with the notch effect of weld concavity, since fatigue is a type of notch-sensitive dynamic failure. As seen in Figs. 5.1(a) and (b), the fiber laser welding

“notch” or weld concavity in both joints was similar to a notched specimen studied by Chapetti *et al.* [88] where the stress concentration was generated, and consequently a decreased fatigue limit or fatigue life at the low level of stress amplitudes was observed. Further studies are needed to eliminate the weld concavity and improve the fatigue resistance.

The obtained fatigue data plotted in Fig. 5.7 could be fitted using the Basquin type equation (4.1). The obtained values of σ_f' and b of the BM, the FLW single linear joint and the FLW multiple linear joints for both HSLA and DP980 steels tested at $R = 0.1$, 50 Hz and room temperature are given in Table 5.2. It is seen from Table 5.2 that after fiber laser welding for both HSLA and DP980 steels the values of b became larger, giving rise to a lower fatigue life after welding.

5.5 Failure Location and Fractography

The typical fatigue failure locations for the HSLA-S joints, the HSLA-M joints, the DP980-S joints and the DP980-M joints are shown in Fig. 5.8. It is seen that the fatigue failure for the FLW HSLA joints occurred in the BM with large deformation at a stress amplitude above 250 MPa (Figs. 5.8(a) and 5.8(b)), whereas it occurred at HAZ below a stress amplitude of 250 MPa (Figs. 5.8(c) and 5.8(d)), irrespective of the single or multiple linear welding. It was believed that while the weld concavity in the FLW HSLA joints (Fig. 5.1(a)) did not significantly influence the tensile strength (Fig. 5.6 and Table 5.2), the fatigue strength at the condition of stress amplitude below 250 MPa was affected adversely. In contrast, the fatigue failure in the FLW DP980 joints always initiated in weld concavity (Figs. 5.8(e) and 5.8(f)) due to the presence of stress concentration, and then propagated in the soft zone (Figs 5.4 and 5.5). The failure location of the FLW joints was similar to that of the DLW DP980 joints in which all the fatigue failure occurred at the soft zones as reported in our earlier study [21].

Fig. 5.8 Typical fatigue failure locations of (a) a HSLA-S joint at a stress amplitude above 250 MPa, (b) a HSLA-M joint at a stress amplitude above 250 MPa, (c) a HSLA-S joint below stress amplitude of 250 MPa, (d) a HSLA-M joint below stress amplitude of 250 MPa, (e) a DP980-S joint, and (f) a DP980-M joint.

SEM images of a fracture surface of a FLW HSLA-S joint tested at an applied stress amplitude of 200 MPa is shown in Fig. 5.9. The cross section in Fig 5.9(a) indicated that the fatigue test sample experienced a large deformation before the final fast crack propagation because of the superior ductility of HSLA steel in comparison with DP980 steel (Table 5.2). It was also confirmed from Fig. 5.9(a) and Fig. 5.9(b) that crack initiation occurred from the weld concavity (Fig. 5.1(a)), corresponding well to the top view of the failed samples (Figs. 5.8(c) and 5.8(d)). A magnified SEM image near the crack initiation site is shown in Fig. 5.9(b), where the crack initiated from the weld concavity could clearly be seen when the applied stress amplitude was below 250 MPa. Fatigue crack propagation was mainly characterized by the formation of fairly typical fatigue striations in conjunction with secondary cracks, as shown in Fig. 5.9(c), while the final fast propagation area consisted of distinctive dimples (Fig. 5.9(d)). Fatigue striations basically occurred through a repeated plastic blunting-sharpening process via the slip of dislocations in the plastic zone in front of the fatigue crack tip [79].

Fig. 5.9 Typical SEM images of fatigue fracture surface of a FLW HSLA-S joint tested at a stress amplitude 200 MPa, (a) overall view, (b) crack initiation area, (c) crack propagation area, and (d) final fast crack propagation area.

SEM images of a fracture surface of a FLW DP980-S joint tested at an applied stress amplitude of 200 MPa is shown in Fig. 5.10. It is from Fig. 5.10(a) that multiple crack initiation occurred from the weld concavity where the surface welding defect occurs (Fig. 5.10(b)) as indicated by the arrow in the image. These characteristics of the fatigue fractograph were similar to those observed on the fatigue fracture surface reported for a FLW AZ31B-H24 Mg alloy [79]. The fatigue crack initiation in the present FLW DP980 steel would primarily be related to the stress concentration caused by the severe weld concavity (Fig. 5.1(b)), in combination with the

presence of detrimental soft zones (Figs 5.4 and 5.5). Like the HSLA welds, fatigue crack propagation in the FLW DP980 joints was also characterized by the occurrence of fatigue striations along with secondary cracks, as shown in Fig. 5.10(c), whereas the final fast propagation area consisted of characteristic dimples as well, as shown in Fig. 5.10(d), where some remaining inclusions located at the bottom of dimples could be seen.

Fig. 5.10 Typical SEM images of fatigue fracture surface of a FLW DP980-S joint tested at a stress amplitude 200 MPa, (a) overall view, (b) crack initiation area, (c) crack propagation area, and (d) final fast crack propagation area.

5.6 Summary

- (1) Single phase microstructure containing martensite formed in the FZ of all the joints; however, newly formed martensite and partially tempered martensite structure was observed in the HAZ of HSLA and DP980 steels, respectively.
- (2) The HSLA joints exhibited a hardness profile with HAZ hardness merging smoothly into the unaffected BM values, while the DP980 joints showed a significant drop of hardness in the HAZ. Despite the formation of fully martensitic structure in the FZ, the higher carbon and manganese content in DP980 steel resulted in a higher FZ hardness (~450 HV) compared to HSLA steel (~360 HV). A characteristic “suspension bridge”-like hardness profile with the FZ hardness as a “pylon” was observed.
- (3) Both the HSLA and DP980 joints showed a superior tensile strength with a joint efficiency reaching about 94~97%. While the ductility of DP980 joints decreased, the ductility of HSLA joints was fairly close to that of BM. Despite the presence of the soft zone, the ultimate tensile strength of the DP980 joints was twice higher than that of the HSLA joints, which corresponded well to the hardness results.
- (4) The DP980 joints had a much longer fatigue life compared to the HSLA joints at a stress amplitude above 250 MPa; however, below this stress amplitude both the DP980 and HSLA joints showed a similar fatigue life.
- (5) No significant effect of weld geometry on the fatigue life of HSLA joints was observed. On the other hand, the fatigue life for the multiple linear DP980 joints exhibited a larger scatter and also lower fatigue strength, indicating that the probability of dynamic fatigue failure at the lower level of stress amplitudes increased with increasing number of soft zone and weld concavity in the weld.
- (6) Above a stress amplitude of 250 MPa, fatigue failure of the HSLA joints occurred in the BM, signifying a superior structural integrity of the fiber laser welded HSLA joints when subjected to higher levels of cyclic loading. When the applied stress amplitude was below 250 MPa, fatigue failure occurred from the weld concavity.
- (7) In the DP980 joints under both higher and lower levels of cyclic loading, fatigue crack was observed to initiate predominantly from the weld concavity and propagate in the soft zone.

Fatigue crack propagation in both HSLA and DP980 joints was characterized by the characteristic fatigue striations coupled with secondary cracks.

CHAPTER 6 MICROSTRUCTURE AND MECHANICAL PROPERTIES OF WELD-BONDED AND RESISTANCE SPOT WELDED MAGNESIUM- TO-STEEL DISSIMILAR JOINTS

6.1 Microstructure

The overall joint nugget and microstructural changes across a WB Mg/Mg joint are shown in Fig. 6.1. It is seen that the cross-section of the weld (Fig. 6.1(a)) exhibited a big microstructural change. At or near the fusion center (Fig. 6.1(b) and (c)) the microstructure consisted of equiaxed dendrites containing divorced eutectic $Mg_{17}Al_{12}$ particles [97-99] mainly located at the interdendritic and intergranular regions. Similar microstructure in the FZ center of AZ31B-H24 Mg alloy after laser welding or double-sided arc welding has been reported in [64,65]. Compared Fig. 6.1(c) with Fig. 6.1(b) it appeared that the dendrite cell size (or grain size) became smaller with increasing distance from the FZ center. Near the FZ border the divorced eutectic $Mg_{17}Al_{12}$ particles were mainly located at the boundaries of less obvious dendrite grains which became further smaller (Fig. 6.1(d)). The microstructural change in the FZ border area can be seen in Fig. 6.1(e), where no divorced eutectic $Mg_{17}Al_{12}$ particles in the HAZ were present. This can be better seen on the image taken in the HAZ (Fig. 6.1(f)). Fig. 1(g) shows the microstructure in the BM (BM) which consisted of deformed and elongated grains, similar to that reported in [64,65]. Even though the grain size (or dendrite cell size) decreased with increasing distance from the FZ center, it was still larger than that of BM, as seen from Fig. 6.1(b-f) vs. Fig. 6.1(g).

Furthermore, it was observed from Fig. 6.1(c-e) that no columnar dendritic structure near the FZ border was developed in the WB Mg/Mg joint. Similar observations of no or undeveloped columnar dendritic structure were also reported in the resistance spot welding by Wang *et al.* [15] on 1 mm thick AZ31B-H24 Mg alloy sheets, and by Xiao *et al.* [100] and Liu *et al.* [101] on the hot-rolled sheets of AZ31 Mg alloy. This was related to the columnar-to-equiaxed transition influenced by second-phase particles in the fusion pool during welding. According Burden and Hunt [102], the columnar-to-equiaxed transition during nugget solidification occurred either by a pile-up of equiaxed crystals that blocked the growth of columnar grains or by the attachment of equiaxed crystals from the liquid to the columnar dendrite front. Xiao *et al.* [100] have recently

observed that during solidification of the fusion pool, when the size of the second-phase particles in front of the interface was large enough, new grains would nucleate and grow in the areas near the FZ boundary and these coarser second-phase particles would block off the epitaxial columnar grains. As a result, columnar dendritic grain growth would be restricted, and equiaxed grains developed, as seen in Fig. 6.1(d) and (e).

Kou [103] presented a constitutional supercooling theory associated with the ratio of temperature gradient (G) to growth rate (R). When G/R ratio is very low, the region of constitutional supercooling is so wide that solid nuclei form in front of the advancing solid/liquid interface and grow into equiaxed dendrites as solidification proceeds. Similar situation was observed by Liu *et al.* [101] who reported that when the welding current increased from 22 kA to 26 kA during RSW of Mg/Mg, the columnar dendritic zone (CDZ) near the FZ border became smaller; when the welding current further increased from 26 kA to 28 kA, the CDZ disappeared. Since the higher welding current gave rise to a higher heat input, leading to a lower temperature gradient and a higher cooling rate which promoted the columnar-to-equiaxed transition. The absence of the columnar structure near the FZ border in the present WB Mg/Mg joint was basically due to more heat generation and higher temperature due to the presence of adhesive added (Fig. 6.2), as reported by [68]. Then the temperature gradient would become lower, resulting in a low G/R ratio and promoting the columnar-to-equiaxed grain transition.

From Fig. 6.1(a) it is also seen that in the WB Mg/Mg joint, solidification or shrinkage cracking occurred. The formation of solidification cracking was basically due to the presence of low melting point liquid films between dendrites due to the segregation of Al and Mn atoms and tensile stress developed during cooling [104,105]. The shrinkage porosity and cavities in the FZ of a given metal were closely related to its thermal expansion in the welding process and forge force at the later stage of welding [106]. Since AZ31 Mg had a relatively high coefficient of volume expansion in liquid at its melting point ($1.65 \times 10^{-4} \text{ K}^{-1}$), in comparison with that of aluminum ($1.31 \times 10^{-4} \text{ K}^{-1}$) or iron ($0.89 \times 10^{-4} \text{ K}^{-1}$), a higher tendency of forming shrinkage porosity and cavities than aluminum or iron would be expected in the fusion welding of Mg/Mg [107].

Fig. 6.1 Microstructures of a WB Mg/Mg joint, (a) overall view of nugget where the location of the subsequent images is indicated, (b) and (c) equiaxed dendritic structure at and near the center of FZ with divorced eutectic $Mg_{17}Al_{12}$ particles, (d) microstructure near the FZ border, (e) FZ/HAZ border area, (f) HAZ, and (g) BM.

Fig. 6.2 A macroscopic image of a typical fatigue fracture surface of the WB Mg/Mg joint, showing the remaining adhesive layer and the central FZ where the adhesive was squeezed or burned off.

Fig. 6.3 and Fig. 6.4 show the microstructures of WB Mg/steel joint and RSW Mg/steel joint, respectively. At the Mg side, both samples had a similar microstructure consisting of four zones, namely, (i) equiaxed dendritic zone (EDZ) in the FZ of Mg alloy close to HSLA in Fig. 6.3(b) and Fig. 6.4(b), (ii) CDZ in Fig. 6.3(d) and Fig. 6.4(d), (iii) HAZ at the top of Fig. 6.3(e) and Fig. 6.4(e), and (iv) BM of AZ31 Mg alloy like Fig. 6.1(g), where Fig. 6.3(c) and Fig. 6.4(c) show the border area between the EDZ and CDZ. It is seen from Fig. 6.4(b) that the Mg FZ adjacent to the HSLA showed some clear shrinkage cracks or cavities in the RSW Mg/steel joint without adhesive, while such shrinkage cracks were basically absent in the WB Mg/steel joint with the addition of adhesive (Fig. 6.3(b)). This would be due to a lower cooling rate in the WB Mg/steel process than in the RSW Mg/steel process. The heat input during resistance spot welding could be expressed by the following equation according to Joule's law [108],

$$Q = I^2 R' t = I^2 \int_0^t R'(t) dt \quad (6.1)$$

where Q is the heat energy in J, I is the welding current in A, t is the welding time in s, and R' is the dynamic resistance in ohm as a function of welding time t . It is clear that both welding current and dynamic resistance can affect the heat generation. However, the welding current has a much stronger effect on the heat generation than the dynamic resistance, i.e., a square dependence vs. a linear dependence. While the adhesive added could increase the dynamic

resistance [68], greater heat generation and higher peak temperature would still be achieved in the present RSW Mg/steel joint, because the constant DC welding current in the RSW Mg/steel process (28 kA, 8 cycles) was higher than that in the WB Mg/steel process (24 kA, 8 cycles with only a slight preheating at 8 kA, 3 cycles). The higher peak temperature in the RSW Mg/steel process would give rise to a faster cooling rate in terms of the thermal theory of convection transfer (sometimes called the Newton's law of cooling) [109,110],

$$\dot{Q} = hA(T_{env} - T(t)) = -hA\Delta T(t) \quad (6.2)$$

where $\dot{Q}$ is the change of thermal energy Q with respect to time t (i.e., dQ/dt), h is the heat transfer coefficient, A is the surface area of the heat being transferred, T is the object's surface temperature, T_{env} is the temperature of the environment. Obviously, the rate of heat loss of a body is proportional to the temperature difference between the body and its surroundings. This has indeed been verified via direct measurements during welding of different materials [7,111,112]. Therefore, a faster cooling rate in the RSW Mg/steel process would be anticipated due to the higher welding current (or heat energy) applied, in comparison with the WB Mg/steel process.

There were essentially no changes in the microstructure at the steel side in WB Mg/steel process with the addition of adhesive due to a slower cooling rate as shown in Fig. 6.3(f) and Fig. 6.3(g) where only ferrite was present in the HAZ. However, in the RSW Mg/steel joint without using any adhesive and with a higher cooling rate, the microstructural change in the HAZ of HSLA occurred to form mainly lath martensite, as compared Fig. 6.4(f) with Fig. 6.4(g). Such a change could be better seen in Fig. 6.5, where more detailed observations in the areas of "h", "i", and "j" indicated in Fig. 6.4(f) via SEM at a higher magnification are presented. As reported in [113,114], the fast cooling rate during RSW of HSLA could lead to the formation of martensite, bainite, pearlite and some retained austenite, depending on the cooling rate, peak temperature, alloying element, etc. Similar studies [21,115] on martensite in a HSLA steel have also indicated that the cooling rates during RSW were very high, which gave an insufficient time for the diffusion of carbon and other alloying elements, leading to the formation of lath martensite containing very thin regions of retained austenite between the laths, or pockets of laths, and possibly some lower bainite, pearlite and some carbides.

It is of interest to observe that nearly no microstructural change in the HAZ of HSLA occurred within approximately 0.1 mm from the edge bordering the magnesium (Fig. 6.4(f)). This could be understood as follows. The heat generation in both steel and magnesium sheets was unequal, which would be higher at the steel side than at the magnesium side because the electrical conductivity of steel was lower than that of magnesium [108,116]. The steel could then serve as a hot anvil to heat the magnesium. Because of the heat losses from both sides of magnesium and steel/electrode with circulating cooling water, only in the center portion of the steel sheet the peak temperature experienced could reach above the austenization temperature (but below the melting point, as indicated by the absence of FZ at the HSLA side). The subsequent rapid cooling led to the formation of martensite alongside some bainite, pearlite, and retained austenite. The presence of the ~0.1 mm non-transformation zone near the interface (Fig. 6.4(f) and Fig. 6.5(a)) indicated that the peak temperature experienced in this area was below the austenization temperature. This might be related to a lower melting temperature of ~630°C for AZ31 Mg alloy, which was below the austenization temperature of HSLA. In addition, heat loss or transfer at the magnesium side would be faster since magnesium had a higher thermal conductivity than steel (i.e., 96 W/m-K [15] for magnesium vs. 27 W/m-K [117] for steel). It follows that the heat loss in the area at the steel side adjoining Mg sheet would also be faster, leading to a lower peak temperature below the austenization temperature and the absence of martensite (Fig. 6.4(f) and Fig. 6.5(a)). In the WB Mg/steel joint, as discussed above, the lower cooling rate prevented the formation of martensite in the HAZ of HSLA (Fig. 6.3(f)), in comparison with the RSW Mg/steel joint (Fig. 6.4(f)).

Fig. 6.3 Microstructures of a WB Mg/steel joint, (a) overall view of nugget where the location of the subsequent images is indicated, (b) equiaxed dendritic structure in the FZ of Mg alloy near HSLA, (c) border area between the equiaxed and columnar dendritic structures, (d) columnar dendritic structure, (e) FZ/HAZ border area, (f) HAZ of HSLA side, and (g) BM of HSLA sheet.

Fig. 6.4 Microstructures of a RSW Mg/steel joint, (a) overall view of nugget where the location of the subsequent images is indicated, (b) equiaxed dendritic structure in the FZ of Mg alloy close to HSLA, (c) border area between the equiaxed and columnar dendritic structures, (d) columnar dendritic structure, (e) FZ/HAZ border area, (f) HAZ of HSLA side, and (g) BM of HSLA sheet.

Fig. 6.5 SEM images showing the microstructural change at the steel side of the RSW Mg/steel joint in Fig.6.4 (f), (a) area h close to the interface, (b) area i, and (c) area j.

6.2 XRD Analysis

XRD analyses were performed on the fracture surface (Mg side) of WB Mg/steel and RSW Mg/steel joints after tensile tests and the obtained results are shown in Fig. 6.6. For each sample, the nugget zone (or FZ) and BM were chosen for XRD scanning. Basically, Mg peaks appeared in the AZ31-H24 BM (Fig. 6.6(a)), while Zn-containing intermetallic compounds (IMCs) were detected on the fracture surface (Fig. 6.6(b,c)) since the HSLA was Zn-coated (hot-dip galvanized) sheet. Based on the binary Mg-Zn phase diagram [118,119], there are five possible intermetallic phases in the Mg-Zn system, namely Mg_7Zn_3 , $MgZn$, Mg_2Zn_3 , $MgZn_2$ and Mg_2Zn_{11} . IMC $MgZn_2$ (also reported in [120,121]) was observed on the fractured nugget zone of the RSW Mg/steel joint as shown in Fig. 6.6(b), while both $MgZn_2$ and Mg_7Zn_3 were detected on

the fractured nugget zone of the WB Mg/steel joint (Fig. 6.6(c)). The Mg_7Zn_3 on the fracture surface of a WB Mg/steel joint could be seen from a typical back-scattered electron SEM image as shown in Fig 6.7(a), which was observed to be present discontinuously in the FZ at the WB Mg/steel joint interface. The corresponding energy EDS analysis (Fig 6.7(b)) revealed an atom percent of 68% and 32% for elements Mg and Zn, respectively, which had an atom ratio close to that of Mg_7Zn_3 , as detected by XRD (Fig. 6.6(c)). It should be noted that the C and O detected on the EDS spectrum were basically from the epoxy ($\text{C}_{21}\text{H}_{25}\text{ClO}_5$) based adhesive.

The formation of IMCs is known to be associated with the diffusion of participating atoms at high temperatures. Since diffusion is a time-dependent process, the formation of additional Mg_7Zn_3 compound was mainly attributed to the lower cooling rate in the WB Mg/steel process. It implied that a longer period of time at relatively higher temperatures would be experienced in the WB Mg/steel process, thus giving rise to enhanced diffusion and promoting the formation of Mg_7Zn_3 compound (Fig. 6.6(c) and Fig. 6.7). Similarly, Wu and Song [122] reported the presence of Mg_7Zn_3 compound in the fracture zone of brazed joints of AZ31B magnesium alloy, and Luo *et al.* [123] observed Mg_7Zn_3 compound present between quasicrystal phases in slowly cooled Mg-Zn-Y alloys. Further studies on the formation mechanisms of Mg_7Zn_3 compound in the WB Mg/steel joint are needed.

(a)

(b)

Fig. 6.6 XRD patterns obtained from the Mg side of fracture surfaces, (a) BM, (b) RSW Mg/steel joint, and (c) WB Mg/steel joint.

(a)

(b)

Fig 6.7 (a) A typical SEM image showing Mg₇Zn₃ formed in the FZ at the WB Mg/steel joint interface, and (b) EDS spectrum of a point analysis in (a).

6.3. Hardness Profiles

Vickers microhardness profiles across the weld nugget (as indicated by the dashed line in Fig. 6.1(a), Fig. 6.3(a) and Fig. 6.4(a)) were determined for the WB Mg/Mg similar joint, WB Mg/steel dissimilar joint, and RSW Mg/steel dissimilar joint, respectively, and the obtained results were plotted in Fig. 6.8. It is seen that in the WB Mg/Mg similar joint a nearly constant hardness value of ~HV60 across the FZ and HAZ was obtained. In the WB Mg/steel dissimilar joint, the Mg side had a similar hardness as that in the WB Mg/Mg similar joint, while the hardness at the steel side reached at an average of HV154, being equivalent to that of the BM. In the RSW Mg/steel dissimilar joint, the hardness of HAZ at steel side showed a gradual increase with increasing distance from the faying surface with a highest value of about HV225 and then decreased. This was obviously attributed to the change in the microstructure as shown in Fig. 6.4(f) and Fig. 6.5. Again, the hardness at the Mg side remained nearly constant at an average of about HV60 as well. For both RSW Mg/steel and WB Mg/steel dissimilar joints, an equivalent value of hardness near the interface between Mg and steel was obtained to be about HV110. Such a small plateau-like feature in both dissimilar joints was due to the presence of intermetallic compounds in the dissimilar welding (Figs. 6.6(b) and 6.6(c)) which resulted in an increase of hardness at the Mg side adjacent to the steel. Besides, the plateau-like feature was somewhat wider for the WB Mg/steel joint than for the RSW Mg/steel joint. This was likely due to the formation of both $MgZn_2$ and Mg_7Zn_3 intermetallic compounds (Fig. 6.6(c) and Fig. 6.7) during weld bonding, which led to a thicker layer of intermetallic compounds in the WB Mg/steel joint. In contrast, there was only $MgZn_2$ formed at the Mg side of RSW Mg/steel joint (Fig. 6.6(b)), thus bringing about a thinner layer intermetallic compound (Fig. 6.8).

Fig. 6.8 Hardness profiles across the WB Mg/Mg similar joint, WB Mg/steel dissimilar joint, and RSW Mg/steel dissimilar joint.

6.4. Tensile Properties

Typical curves of the applied load versus displacement under three welding conditions are shown in Fig. 6.9. It is seen that both WB joints were much stronger than RSW Mg/steel joint, while WB Mg/Mg joint was further stronger than WB Mg/steel joint. The evaluated maximum tensile shear load is shown in Fig. 6.10(a). It is clear that the maximum tensile shear load of the WB Mg/Mg joint was higher than that of the WB Mg/steel joint, which was further higher than that of the RSW Mg/steel joint. While the maximum tensile shear load of the RSW Mg/steel joint decreased slightly with increasing crosshead speed, the maximum tensile shear load of both the WB Mg/Mg and WB Mg/steel joints increased slightly. All of the tensile shear tested samples showed an interfacial fracture mode (Fig. 6.2).

Fig. 6.9 Tensile load versus displacement for the WB Mg/Mg similar joint, WB Mg/steel dissimilar joint, and RSW Mg/steel dissimilar joint.

To characterize more completely the tensile shear characteristics of the welded joints and better identify the effect of welding processes, energy absorption has been estimated, where the energy absorption is defined as the area below the load-displacement curve (Fig. 6.9). When the tensile shear remained continuous after reaching the maximum load without failure (i.e., in the cases of WB Mg/Mg and WB Mg/steel joints in the present study), the energy absorption is calculated corresponding to a 5% load drop relative to the maximum load. The evaluated results are plotted in Fig. 6.10(b). The energy absorption of both the WB Mg/Mg and WB Mg/steel joints was substantially higher than that of the RSW Mg/steel joint, while the energy absorption of the WB Mg/Mg joint was slightly higher than that of WB Mg/steel joint. Similar to the maximum shear load (Fig. 6.10(a)), the energy absorption of the RSW Mg/steel decreased with increasing strain rate (or crosshead speed). This was basically due to adiabatic heating during the tensile test [124,125].

On the other hand, the absorbed energy of WB Mg/Mg and WB Mg/steel joints increased to a greater extent with increasing crosshead speed. This reflects the beneficial role of weld adhesive,

since the energy absorption capability is an important parameter in vehicle crashworthiness. There exists a direct relationship between the failure energy in the static tensile shear tests and impact tests. This implies that WB welding had a significantly higher impact resistance than RSW welding. Similar variation trend was also reported in [126-129]. For instance, Challita *et al.* [128] observed that the highest stress increased with increasing crosshead speed in the tensile test of the double-lap adhesively bonding. Goglio *et al.* [129] studied the effect of strain rate on the mechanical behavior of epoxy adhesives and observed that the maximum stress of the tested adhesive was very sensitive to the strain rate (i.e., increased with increasing strain rate). The addition of adhesive in the present WB joints would effectively promote the positive strain rate dependence. The significant improvement in both the maximum tensile shear load and energy absorption (Figs. 6.10(a) and 6.10(b)) suggests that the added adhesive played a predominant role in the joints. The superposition of adhesive bonding and spot welding led to the maximum load and especially the energy absorption increasing with increasing crosshead speed.

(a)

Fig. 6.10 (a) The maximum tensile shear load and (b) the energy absorption versus crosshead speed for the WB Mg/Mg similar joint, WB Mg/steel dissimilar joint, and RSW Mg/steel dissimilar joint.

6.5 Fatigue Strength and Failure Mode

Fatigue test results of WB Mg/Mg, WB Mg/steel and RSW Mg/steel joints obtained at RT, $R=0.2$, and 50 Hz were plotted in Fig. 6.11. The fatigue life of both the WB Mg/Mg and WB Mg/steel joints was equivalent within the experimental scatter when the applied maximum cyclic load was below about 10 kN, which was much longer than that of the RSW Mg/steel joint at a given maximum cyclic load. Indeed, the fatigue strength (reflected by the maximum cyclic load in Fig. 6.11) at a given number of cycles to failure was about three times higher for the WB Mg/Mg and WB Mg/steel joints than for the RSW Mg/steel joint. For example, at a fatigue life of $\sim 1 \times 10^5$ cycles, it is seen that the fatigue strength of the WB Mg/Mg and WB Mg/steel joints was about three-fold higher than that of the RSW Mg/steel joint. The greatly improved fatigue resistance of the WB Mg/Mg and WB Mg/steel joints was primarily attributed to the weakening or elimination of stress concentration surrounding the weld nugget/button as a result of large bonded area and uniform stress distribution in the joint during fatigue tests. Similar results were

reported in [127,130,131], where the fatigue life of the WB joint was longer than that of the RSW joint in the advanced high strength steels or aluminum alloy 5754-H40 as well. Therefore, the WB joints exhibited a significant improvement over the RSW joints. The equivalency of the S-N type of curves between the WB Mg/Mg similar joint and WB Mg/steel dissimilar joint when the maximum cyclic load was lower than ~10 kN in Fig. 6.11 indeed reflected the fatigue life of Mg base alloy itself, since the failure of all tested samples exclusively occurred in the BM at the Mg side and normal to the loading direction, as shown in Figs. 6.12(a) and 6.12(b). In contrast, the failure of RSW Mg/steel joint occurred in the HAZ, i.e., the fatigue crack initiated from the HAZ and then propagated perpendicular to the loading direction at the maximum cyclic load of below 3 kN, as seen in Fig. 6.12(c), and in the mode of interfacial fracture at the maximum cyclic loads ranging in-between 3~6 kN (Fig. 6.12(d)). These observations indicated again that the WB joints were robust and durable without fatigue failure occurred at the joints even up to a maximum cyclic load of 10 kN, reflecting their high reliability and integrity. When the applied maximum cyclic load was above ~10 kN, while the WB Mg/Mg similar joints still exhibited a nearly linear change up to a maximum cyclic load as high as 16 kN, there existed an obvious inflection on the curve of WB Mg/steel dissimilar joints (Fig. 6.11), indicating a rapid reduction of fatigue life with increasing maximum cyclic load. This was due to the occurrence of large plastic deformation at the BM of HSLA side, as shown in Fig. 6.13, because the maximum cyclic stress applied on the 0.7 mm thick base steel had exceeded its yield strength of HSLA. It should be pointed out that out of the four fatigue samples of WB Mg/steel dissimilar joints tested above the maximum cyclic load of 10 kN, three of them failed at the BM of HSLA side due to the large plastic deformation shown in Fig. 6.13, while one of them failed in the mode of interfacial fracture which will be shown later. This implied that there was still remaining potential for the WB Mg/steel joint itself to endure a higher level of cyclic loading. To let such a WB Mg/steel dissimilar joint exert its full load-carrying capacity, either a thicker HSLA sheet or a higher grade of steel could be weld-bonded to the 2 mm thick AZ31 Mg alloy. Further studies in this aspect are needed. For the WB Mg/Mg joints, while the majority of samples tested at a maximum cyclic load below 16 kN failed in the BM at the Mg side (Fig. 6.12(a)), partial button pullout failure was observed when the applied maximum cyclic load reached 16 kN, as seen from a three-dimensional image shown in Fig. 6.14. As mentioned above and shown in Fig. 6.12, the crack initiation sites for the WB Mg/steel joint and the RSW Mg/steel joint were different. While

the RSW Mg/steel joints failed either due to crack initiation from the HAZ at the notch root and then crack propagation along the direction perpendicular to the loading direction (Fig. 6.12(c)) or in the mode of interfacial fracture (Fig. 6.12(d)), it should be noted that the notch root around the spot weld nugget was a place where the triaxial stresses reached the maximum value due to both the maximum shear stress and normal stress arising from the bending moment during lap shear tensile test [17,132]. Fatigue crack basically initiated at the edge of the overlapping/bonding area in the BM of Mg side for the WB Mg/steel joints (Fig. 6.12(b)). This was understandable since a certain degree of stress concentration was still present at the edge of bonding area, although the stress concentration around the spot weld button was effectively reduced or eradicated by the adhesive bonding. Similar observations have also been reported in [133,134].

Fig. 6.11 S-N curves of the WB Mg/Mg similar joint, WB Mg/steel dissimilar joint, and RSW Mg/steel dissimilar joint tested at RT, $R=0.2$, and a frequency of 50 Hz.

Fig. 6.12 Fatigue failure modes in (a) WB Mg/Mg joint at a maximum cyclic load of 12 kN, (b) WB Mg/steel joint at a maximum cyclic load of 10 kN, (c) RSW Mg/steel joint at a maximum cyclic load of 2 kN, and (d) RSW Mg/steel joint at a maximum cyclic load of 3 kN.

Fig. 6.13 Fatigue failure location of the WB Mg/steel joint at an applied maximum cyclic load of 12 kN.

Fig. 6.14 A typical 3D SEM image showing the partial button pullout of the WB Mg/Mg similar joint after fatigue test at an applied cyclic maximum load of 16 kN.

6.6 Fractography

To observe more details on the fracture surface, SEM images of the partial button pull-out failure of a WB Mg/Mg similar joint tested at a maximum cyclic load of 16 kN are shown in Fig. 6.15, where the cracks initiated from both sites “b” and “d” and grew across the weld nugget with increasing number of cycles. The feature for the crack initiation is shown in Fig. 6.15(b), which was fairly similar to that occurred in a typical RSW joint where the triaxial stresses reached the maximum value during the lap shear tensile tests. In addition, the internal shrinkage cracks and porosity in the weld nugget were also observed on the fractograph (Figs. 6.15(c) and 6.15(d)). Fatigue crack could be readily developed by growing and coalescing these shrinkage cracks and cavities located in the middle of the spot weld nugget, where the irregular shape of the cavities was very conducive to crack initiation and propagation with little plastic deformation. Fig. 6.15(d) shows clearly the dendritic structure in a cavity. The region in the cavity contained dendrite lobes, which existed as a “free surface” within the cavity after available liquid metal solidified. This feature might have been caused by a lack of molten metal within the cavity [17,36,131,]. The typical crack propagation zone is shown in Fig. 6.15(c) for area “c” at a lower magnification and in Fig. 6.15(e) for area “e” with elliptical shear deformation characteristics.

The dimple features in Fig. 6.15(f) indicated that the localized tensile stress appeared at the top of button pull-out part (Fig. 6.15(a)).

Fig. 6.15 SEM images of a fatigue fracture surface of a WB Mg/Mg similar joint tested at an applied cyclic maximum load of 16 kN, (a) overall view, (b) near the crack initiation site, (c) shrinkage crack in FZ, (d) shrinkage pore at a higher magnification, (e) facet-like shear deformation features, and (f) dimple features at the top of pull-out button.

Fig. 6.16 presents SEM images of fatigue fracture surfaces of the WB Mg/steel and RSW Mg/steel dissimilar joints failed in the mode of interfacial fracture. At the Mg side of the WB Mg/steel joint (Fig. 6.16(a)) the overall fracture surface appeared smoother than that of the RSW Mg/steel joint (Fig. 6.16(b)). This was likely due to the presence of additional intermetallic compound of Mg_7Zn_3 , as identified via XRD (Fig. 6.6(c) and Fig. 6.7). As mentioned earlier, in the WB Mg/steel joints tested above a maximum cyclic load of 10 kN (Fig. 6.11), while three samples failed at the BM of HSLA side due to the large plastic deformation (Fig. 6.13), one sample failed in the mode of interfacial fracture (Fig. 6.16(a)). This was related to the presence of a visible initial shrinkage crack of ~1.5 mm long located near the top of the joint where the maximum stress also occurred (Fig. 6.16(a)), thus causing premature fatigue crack initiation from this location in this specific specimen. In view of the absence of interfacial failure in all other samples tested over a wide range of maximum cyclic loads from 3 kN to 12 kN (i.e., three samples above 10 kN and all other samples at or below 10 kN, as shown in Fig. 6.11), no serious defects in the weld nugget of the WB Mg/steel dissimilar joints would be expected. Indeed, by comparing the FZ of the WB Mg/steel joint with that of RSW Mg/steel joint in Fig. 6.16(c) and Fig. 6.16(d), smaller shrinkage cracks and cavities occurred more frequently in the RSW Mg/steel joint, which corresponded well to the cross-sectional observations shown in Fig. 6.3(b) and Fig. 6.4(b). This suggests that, despite the formation of additional intermetallic compound of Mg_7Zn_3 , the lower cooling rate in the WB Mg/steel joints could generally reduce the tendency for the formation of shrinkage cracks and cavities, thus leading to improved quality of joints, in addition to the prevalent bonding role in reducing or eliminating the stress concentration around the spot weld nugget.

Fig. 6.16 SEM images of fatigue fracture surfaces of the WB Mg/steel and RSW Mg/steel dissimilar joints taken from the Mg side, (a) interfacial fracture surface of the WB Mg/steel joint tested at an applied cyclic maximum load of 11 kN, (b) interfacial fracture surface of the RSW Mg/steel joint tested at an applied cyclic maximum load of 5 kN, (c) FZ of the WB Mg/steel joint, and (d) FZ of the RSW Mg/steel joint.

6.7 Summary

- (1) In the WB Mg/Mg similar joint, typical equiaxed dendritic and divorced eutectic structures formed in the FZ. No columnar dendritic structures near the fusion boundary were observed. This indicated that the applied adhesive layer reduced the temperature gradient and promoted columnar-to-equiaxed transition during the weld-bonding process. The hardness across the WB Mg/Mg similar joint remained basically constant.
- (2) In the dissimilar joints of RSW and WB Mg/steel, FZ appeared only at the Mg side, consisting of equiaxed dendrites adjacent to the steel, and columnar dendrites near the fusion

boundary. Compared with the RSW Mg/steel joint, the lower cooling rate in the WB Mg/steel joint tended to impede the presence of shrinkage cracks and cavities. At the steel side the microstructure did not change in the WB Mg/steel joint due to the lower cooling rate while it changed from polygonal ferrite to a mixture of lath martensite, bainite, pearlite and retained austenite, leading to a significant increase in the microhardness in the RSW Mg/steel joint.

- (3) X-ray diffraction revealed the formation of MgZn_2 in the RSW Mg/steel joint due to the presence of zinc coating on the hot-dip galvanized HSLA steel. The applied adhesive layer led to the occurrence of both MgZn_2 and Mg_7Zn_3 in the WB Mg/steel joint. This was due to the fact that the lower cooling rate in the weld bonding process equivalently increased diffusion time and thus promoted the formation of Mg_7Zn_3 .
- (4) Tensile shear tests showed that the maximum tensile shear load was the highest for the WB Mg/Mg joint and the lowest for the RSW Mg/steel joint, with the values of the WB Mg/steel joint lying in-between them. However, the energy absorption of both WB Mg/Mg and WB Mg/steel joints was equivalent, which was substantially higher than that of the RSW Mg/steel joint.
- (5) While the maximum tensile shear load and energy absorption decreased slightly with increasing crosshead speed (or strain rate) in the RSW Mg/steel joint, they increased with increasing strain rate in both WB joints of Mg/Mg and Mg/steel. This reflected a beneficial role of weld adhesive in enhancing the energy absorption capacity in vehicle crashworthiness.
- (6) Fatigue life of both the WB Mg/Mg and WB Mg/steel joints was equivalent when the applied maximum cyclic load was below ~ 10 kN, which was much longer than that of the RSW Mg/steel joint. The fatigue strength at a given number of cycles was about three times higher for the WB Mg/Mg and WB Mg/steel joints than for the RSW Mg/steel joint. This was primarily attributed to the weakened stress concentration around the spot weld nugget as a result of large bonded area and more uniform stress distribution in the joint.
- (7) In the RSW Mg/steel joint fatigue crack initiated from the HAZ at the notch root and propagated perpendicular to the loading direction at the maximum cyclic load below 3 kN, and fatigue failure occurred in the mode of interfacial fracture at the maximum cyclic load above 3 kN. The failure of both WB Mg/Mg and WB Mg/steel joints basically occurred in

the BM at the Mg side and normal to the loading direction at a maximum cyclic load below ~10 kN, above which plastic deformation and final failure occurred mostly at the BM of steel side in the WB Mg/steel joint as well as partial button pull-out occurred in the WB Mg/Mg joint.

CHAPTER 7 MICROSTRUCTURE AND MECHANICAL PROPERTIES OF WELD-BONDED AND ADHESIVE-BONDED MAGNESIUM JOINTS

7.1 Microstructure

The overall weld nugget and microstructure of WB-1 Mg/Mg joint are shown in Fig. 7.1. The cross-section of the weld (Fig. 7.1(a)) experienced a microstructural change as reported in Chapter 6. The microstructure of the weld nugget consisted of equiaxed dendrites containing divorced eutectic $\text{Mg}_{17}\text{Al}_{12}$ particles [97-99], which were mainly located at the interdendritic and intergranular regions (Fig. 7.1(b)). Little divorced eutectic $\text{Mg}_{17}\text{Al}_{12}$ particles in the heat-affected zone (HAZ) were observed, as shown in Fig. 7.1(c) and (d), where HAZ was characterized by equiaxed recrystallized grains. Fig. 7.1(e) shows the microstructure in the BM AZ31B-H24 Mg alloy which consisted of deformed and elongated grains, similar to those reported in [64,65].

The weld nugget and microstructural changes across a WB-0.5 joint are shown in Fig. 7.2. The microstructure in the nugget of the WB-0.5 joint also consisted of equiaxed dendrites and $\text{Mg}_{17}\text{Al}_{12}$ particles located at the interdendritic and intergranular regions as seen from Fig. 7.2(a); no or little divorced eutectic $\text{Mg}_{17}\text{Al}_{12}$ particles in the HAZ of the WB-0.5 joint were present, as shown in Fig. 7.2(b) and 7.2(c). By comparing the WB-1 with WB-0.5 joints, less solidification cracking or shrinkage was observed in the WB-0.5 joints. This might be due to a slower cooling rate in the cooling phase after RSW when reducing the bonding overlap area. It is known that most of the heat was diffused by cooling water in the electrodes during RSW. Other means of heat loss/transfer would be through Mg alloy itself and the surrounding air. According to the following *Fourier's law* of heat conduction [135],

$$\bar{q} = -k\nabla T, \quad (7.1)$$

where $\bar{q}$ is the local heat flux in $\text{W}/(\text{m}^2)$, k is the material's conductivity in $\text{W}/(\text{m}\cdot\text{K})$, ∇T is the temperature gradient in K/m . Since the thermal conductivity of Mg (about 10^2 to 10^3 $\text{W}/\text{m}\cdot\text{K}$) was much higher than that of air (about 10^{-4} to 10^{-6} $\text{W}/\text{m}\cdot\text{K}$) [136], the heat transfer by Mg workpieces would be more important in a weld bonding process. A smaller overlap area of the WB-0.5 joint represented a smaller or less efficient heat transfer area and a lower heat storage

volume or capacity. This would result in a lower cooling rate in the WB-0.5 joint, which was helpful for reducing pores and hot cracks [105,137]. On the other hand, when the overlap bonding area was larger, it was hard to squeeze the adhesive out of the fusion zone because of the poor liquidity of the adhesive. Once adhesives were trapped in the nugget, pores could be induced since the adhesives were made by epoxy and could gasify when heating up. Therefore, WB-0.5 joints exhibited fewer pores in the nugget than WB-1 joints, as shown in Fig. 7.2(a) vs. Fig. 7.1(a).

Fig. 7.1 Microstructures of a WB-1 Mg/Mg joint, (a) overall view of nugget where the location of the subsequent images is indicated, (b) equiaxed dendritic structure (c) microstructure near the FZ border, (d) HAZ, and (e) BM (BM).

Fig. 7.2 Microstructures of a WB-0.5 Mg/Mg joint, (a) overall view of nugget where the location of the subsequent images is indicated, (b) equiaxed dendritic structure (c) microstructure near the FZ border, (d) HAZ, and (e) BM (BM).

7.2 Tensile Properties

Typical curves of the applied tensile shear load versus displacement for WB-1, WB-0.5 and AB-0.5 joints are shown in Fig. 7.3. The WB-1 joint was stronger than both WB-0.5 and AB-0.5 joints because of the larger bonding area, while the maximum load of the WB-0.5 joint was slightly lower than that of the AB-0.5 joint. This might be due to the fact that the spot weld of WB joints would have some less favorable effect on the stress distribution during tensile shear

tests as reported by Chang *et al* [138]. based on both their experimental and numerical results on a 08Al steel sheet (a car body steel sheet) and SHA-2858 II epoxy-based adhesives, where the welding heat and welding force destroyed/squeezed the adhesive layer and resulted in a reduction in the actual loading area and an increase in both normal stress (within the weld nugget) and shear stress (near the border of the nugget) [138]. Similar stress distributions in the weld nugget across a WB joint via a finite element analysis were reported in [133].

Fig. 7.3 Tensile shear load versus displacement for the WB-1 Mg/Mg joint, WB-0.5 Mg/Mg joint and AB-0.5 Mg/Mg joint tested at a crosshead speed of 0.1 mm/min.

The evaluated maximum tensile shear load (failure load) is shown in Fig. 7.4(a). It is seen that maximum tensile shear loads of WB-1, WB-0.5 and AB-0.5 joints increased with increasing crosshead speed. A similar trend in AB joints was reported in references [71,128,129], For instance, Challita *et al.* [128] observed that the highest stress increased with increasing crosshead speed during a tensile shear test of double-lap adhesively bonding. Goglio *et al.* [129] also

reported that the maximum stress of the tested adhesive was very sensitive to the strain rate (i.e., increased with increasing strain rate).

To better understand the tensile shear characteristics of the three types of joints and identify the effect of adhesive bonding length, the maximum tensile shear stress (i.e., the maximum tensile load divided by the overlap area) versus crosshead speed has been estimated. As shown in Fig 7.4(b), the WB-0.5 joints had a higher maximum tensile shear stress than the WB-1 joints, while it was slightly lower than that of AB-0.5 joints. A similar result was reported by Liu & Ren [139] by comparing a metal inert gas (MIG) spot weld bonding with an adhesive bonding on AZ31 Mg alloy and 6061 Al alloy sheets, where the tensile shear test showed that the strength of MIG spot weld bonding (maximum tensile shear load: 5.3 kN) was slightly lower than that of the adhesive bonding (maximum tensile shear load: 6.0 kN), and both were much higher than that of the MIG spot welding (maximum tensile shear load: 1.7 kN).

(a)

(b)

(c)

Fig. 7.4 (a) The maximum tensile shear load, (b) the maximum tensile shear stress, and (c) the energy absorption versus the crosshead speed for the WB-1 Mg/Mg, WB-0.5 Mg/Mg and AB-0.5 Mg/Mg joints.

The energy absorption (i.e., the area below a load-displacement curve) was calculated according to the curves in Fig. 7.3. As seen from Fig. 7.4(c), with increasing crosshead speeds the energy absorption of the WB-1 joints obviously increased, while the absorbed energy of WB-0.5 and AB-0.5 joints kept nearly constant and was almost equivalent.

7.3 Fatigue Strength and Failure Mode

Fatigue test results obtained at RT, $R=0.2$, and a frequency of 50 Hz are plotted in Fig. 7.5. It is seen that at the same maximum load level, the fatigue life of the WB-1 joints was longer than that of both WB-0.5 and AB-0.5 joints due to a longer bonding. The fatigue life of the WB-0.5 joints was shorter than that of the AB-0.5 joints at the same maximum load level. Similar results on the fatigue life of 08Al steel joints with SHA-2858 II epoxy-based adhesives were reported by Chang *et al.* [138], where the adhesive-bonded joints had a longer fatigue life than the weld-bonded joints at all cyclic loading levels applied. As discussed above, the presence of the unbonded area (i.e., the plastic ring at the peripheral area of the nugget) induced by the electrode force could increase the local stress at both the notch of the weld nugget and the edge of the adhered overlap [133,138,]. The higher stress made the fatigue crack initiation and growth from the weld cavities easier [131,133,138]. Fig. 7.5(b) presented the maximum tensile shear stress versus the number of cycles to failure of WB-1, WB-0.5 and AB-0.5 joints. It is seen that the AB-0.5 joints showed the highest fatigue resistance because of a uniform stress distribution. Both WB-1 and WB-0.5 joints performed equivalently with respect of fatigue life at low cycle fatigue regime (~ 12 MPa). However, at high cycle fatigue region (or lower stress amplitude levels), WB-0.5 joints showed a longer fatigue life than WB-1 joints because of the fewer solidification or shrinkage cracks in the nugget in WB-0.5 joints as shown in Fig. 7.1(a). This corresponded well to that presented by Roesler *et al.* [86] who pointed out that the fatigue strength of a material under dynamic cyclical loading was much more susceptible to the manufacturing process and materials than the static strength. The fatigue strength was also much more sensitive to the lower level cyclic load than the higher level cyclic load.

Fig. 7.6 presents the fatigue data in conjunction with the failure mode by plotting the maximum load vs. the number of reversals to failure ($2N_f$). It should be noted that the run-out data for the non-failed samples at or over 1×10^7 cycles were not included in the fitting. For all the WB-1, WB-0.5 and AB-0.5 joints tested, cohesive failure mode along the adhesive layer was observed at the higher maximum load levels, and BM failure occurred at the lower maximum load levels (Fig. 7.6). When the cohesive failure occurred, partial nugget pull-out from the weld could be observed in the WB-0.5 and WB-1 joints, as shown in Fig. 7.7 (a maximum cyclic load of 16 kN for a WB-0.5 joint). The change of the failure mode from the cohesive failure to the BM failure occurred when $\text{Log}P_{\text{max}}=1.21$ for WB-1 joints (Fig. 7.6(a)), which was much higher than that of the WB-0.5 joints ($\text{Log}P_{\text{max}}=0.81$) (Fig. 7.6(b)). This suggests a higher fatigue resistance of WB-1 at a higher cyclic load level. Similarly, when $\text{Log}P_{\text{max}}=0.93$, this change occurred for the AB-0.5 joints (Fig. 7.6(c)).

Fig. 7.6 The maximum load vs. the number of reversals to failure ($2N_f$) in the double-log scale for (a) WB-1 Mg/Mg joints, (b) WB-0.5 Mg/Mg joints, and (c) AB-0.5 Mg/Mg joints.

Fig. 7.7 A typical 3D SEM image showing the partial button pullout of the WB-0.5 Mg/Mg joint after fatigue test at an applied cyclic maximum load of 16 kN.

7.4 Fractography

To observe more details on the fracture surface, SEM images of the failed WB-1 and WB-0.5 joints are shown in Figs 7.8 and 7.9 which were tested at a maximum cyclic load of 16 kN and 10 kN, respectively. It can be seen in Fig. 7.8(a) that fatigue crack could be developed from the point “d” and propagated along the faying surface and through the nugget. The typical elliptical shear deformation characteristics as shown in Fig. 7.8(b) and 7.8(c) reflected that shear stress dominated both fatigue crack initiation, propagation and final rapid failure area. Free solidified dendritic structure was observed in the center of the nugget, as shown in Fig. 7.8(d). The shrinkage pores in the nugget of the WB-1 joint (Fig. 7.8(a)) were more than that of the WB-0.5 joint (Fig. 7.9(a)), which matched well with the cross-section observations shown in Fig. 7.1(a) and Fig. 7.2(a). This would be the reason why a higher fatigue resistance or a longer fatigue life at a given stress maximum cyclic stress was observed in the WB-0.5 joints, as shown in Fig. 7.5(b). Except the above difference, the fracture surface of the WB-0.5 joint displayed basically a similar feature to that of the WB-1 joint, such as the typical elliptical shear deformation characteristics (Fig. 7.9(b) and 7.9(c)) and free solidified dendritic structure (Fig. 7.9(d)).

Fig. 7.8 SEM images of a fatigue fracture surface of a WB-1 Mg/Mg joint tested at an applied cyclic maximum load of 16 kN, (a) overall view, (b) facet-like shear deformation features near the interfacial surface, (c) facet-like shear deformation features at the top of partial pull-out button, and (d) shrinkage pore in the weld nugget.

Fig. 7.9 SEM images of a fatigue fracture surface of a WB-0.5 Mg/Mg joint tested at an applied cyclic maximum load of 10 kN, (a) overall view, (b) facet-like shear deformation features near the interfacial surface, (c) facet-like shear deformation features at the top of partial pull-out button, and (d) shrinkage pore in the weld nugget.

7.5 Summary

- (1) In both WB-1 Mg/Mg joints with a larger bonding area of 35mm×35mm and WB-0.5 Mg/Mg joints with a smaller (half) bonding area of 17.5mm×35mm, typical equiaxed dendritic structures containing divorced eutectic $Mg_{17}Al_{12}$ particles formed in the FZ, while less severe solidification shrinkage pores were observed to be present in the WB-0.5 joints. The HAZ in both types of joints was characterized by equiaxed recrystallized grains.
- (2) Although the WB-0.5 joints exhibited a slightly lower maximum tensile shear stress than the adhesive-bonded AB-0.5 joints with a bonding area of 17.5mm×35mm due to the reduction

of actual loading area and the increase in the stress across the weld nugget in the WB-0.5 joints, the energy absorption was equivalent. However, the energy absorption of WB-1 joints was much higher due to the larger bonding area, which also increased with increasing crosshead speed.

- (3) The AB-0.5 joints exhibited a higher fatigue resistance at higher cyclic stress levels because of the presence of a uniform stress distribution in the bonding area, while the WB-0.5 joints showed an equivalent fatigue resistance at lower cyclic stress levels. Also, a higher fatigue resistance was observed for the WB-0.5 joints compared with the WB-1 joints due to the occurrences of fewer solidification shrinkage pores.
- (4) In all tested joints, cohesive failure mode along the adhesive layer was observed at the higher cyclic load levels, and fatigue failure occurred in the base metal at the lower cyclic load levels. The cohesive failure was often accompanied by the occurrence of partial nugget pull-out from the weld.

CHAPTER 8 THE TEMPERATURE DEPENDENCE OF THE TENSILE PROPERTIES OF ADHESIVE-BONDED AND WELD-BONDED JOINTS

8.1 Microstructure

Fig. 8.1 showed a typical cross section of the bonding center in an AB-0.5 Mg/Mg joint, and the thickness of cured adhesive layer between two workpieces was to be about 50 μm . The AB-0.5 Mg/Mg joint had no any microstructure change during the curing process because the recrystallization temperature of Mg AZ31B-H24 alloy (about 250°C) is higher than the adhesive curing temperature at 180°C [140,141].

Fig. 8.1 Microstructure of an AB-0.5 Mg/Mg joint.

The detailed results and discussion of microstructures in the WB-1 Mg/Mg, WB Mg/steel, RSW Mg/steel, WB-0.5 Mg/Mg and AB-0.5 Mg/Mg joints were given in Chapters 6 & 7. SEM photographs in Fig. 8.2 showed the microstructure of a WB-1 Mg/Mg joint (Fig. 8.2(a)) and a WB-0.5 Mg/Mg joint (Fig. 8.2(b)), respectively. The former showed severe solidification shrinkage cracks or pores. As noted in chapter 7, half an overlap area for the WB-0.5 Mg/Mg joint represented less efficient heat transfer area and lower heat storage volume/capacity which resulted in a lower cooling rate in comparison with the WB-1 Mg/Mg joint. This was the main reason why the WB-0.5 Mg/Mg joint reached higher maximum tensile shear stress and fatigue resistance compared with the WB-1 Mg/Mg joint even though the bonding area reduced 50%.

Fig. 8.2 Microstructure of (a) WB-1 Mg/Mg joint, and (b) WB-0.5 Mg/Mg joint.

Fig. 8.3 shows the SEM micrograph of welding nugget in a RSW Mg/steel joint (Fig. 8.3(a)) and a WB Mg/steel joint (Fig. 8.3(b)). It was observed that the RSW Mg/steel joint had obvious solidification shrinkage cracks or pores than the WB Mg/steel joint because of faster cooling rate. This observation plus adhesive layer resulted in the maximum tensile shear load of the WB Mg/steel joint to be much higher than that of the RSW Mg/steel joint as discussed in chapter 6. Similarly, when the applied maximum cyclic load was below ~ 10 kN the fatigue strength at a given number of cycles was about three times higher for the WB Mg/steel joints than for the RSW Mg/steel joints.

Fig. 8.3 Microstructure of (a) RSW Mg/steel joint, and (b) WB Mg/steel joint.

8.2 Tensile Properties and Energy Absorption

Typical curves of the applied load versus the displacement at different temperatures under four bonding conditions are shown in Fig. 8.4. It is seen that the AB-0.5 Mg/Mg joints became weaker (maximum tensile load: 7.3 kN) at high temperature of 90°C than that (maximum tensile load: 14.4 kN) at room temperature because the adhesive had lost some stiffness, but it was much more ductile at high temperature [76]. Likewise, Adamvalli & Parameswarran [142] reported that the maximum tensile load at 100°C was 18% less than that at 25°C for a Ti-Al-4V alloy lap joint by using Araldite 2014 (Huntsman) epoxy-based structural adhesive during a tensile test. The AB-0.5 Mg/Mg joints (maximum tensile load: 15.2 kN) at a lower temperature were stronger than that (maximum tensile load: 14.4 kN) at room temperature (Fig. 8.4(a)). This was because the adhesive was stiffer and more brittle, and slightly stronger at a lower environmental temperature [76]. A similar result was reported by Harris & Fay [143] by using a mild steel confirming to BS1449: CRIE (British Standards Institution) and XW1012 epoxy adhesive supplied by Ciba-Geigy using a tensile test at a temperature of -30°C.

Fig. 8.4(b) shows the tensile test results of the WB-0.5 Mg/Mg joints at varying test temperatures. It is seen that the maximum tensile shear loads for the WB-0.5 Mg/Mg joints experienced a similar change to that of the AB-0.5 Mg/Mg joints. With respect to the effect of temperatures, the WB-0.5 Mg/Mg joints (maximum tensile load: 7.7 kN) were weaker at a higher temperature than that (maximum tensile load: 12.4 kN) at room temperature, this was a direct result of the adhesive softening upon heating (as do most viscoelastic materials), thereby having a reduced ability to carry the tensile shear loads at elevated temperatures (i.e., 90°C) [76]. However, it (maximum tensile load: 12.9 kN) becomes stronger at a lower temperature than that (maximum tensile load: 12.4 kN) at room temperature because the YS increased at the lower temperature (i.e., -40°C) for both Mg and steel alloy as well as the epoxy adhesive [74]. Furthermore, in this case, spot weld of WB joint might have a negative effect on the stress distribution during tensile shear test due to the fact that welding heat destroyed the adhesive layer and resulted in a reduction of actual loading area and the stresses in the lap region were increased at room temperature [138].

(a)

(b)

(c)

(d)

Fig. 8.4 Typical tensile shear load versus displacement curves of (a) AB-0.5 Mg/Mg joint, (b) WB-0.5 Mg/Mg joint, (c) WB-1 Mg/Mg joint and (d) WB Mg/steel joint (LT: low temperature, RM: room temperature, and HT: high temperature).

The WB-1 Mg/Mg and WB Mg/steel joints also exhibited a similar trend as shown in Figs 8.4(c) and 8.4(d), respectively. Both joints became weaker at the higher temperature (maximum tensile load: 11.8 kN and 9.3 kN) than at room temperature (maximum tensile load: 18.8 kN and 11.0 kN). However, the WB-1 Mg/Mg and WB Mg/steel joints were stronger at the lower temperature (maximum tensile load: 20.0 kN and 11.8 kN) than at room temperature (maximum tensile load: 18.8 kN and 11.0 kN).

The evaluated maximum tensile shear load versus temperature for the AB-0.5 Mg/Mg, WB-0.5 Mg/Mg and WB-1 Mg/Mg joints is shown in Fig. 8.5(a). It is clear that the maximum tensile shear load for all joints increased with decreasing temperature. Similar variation trend for AB and WB joints was reported in [74-76]. Similarly, the evaluated maximum tensile shear stress versus temperature for above joints is shown in Fig. 8.5(b). The AB-0.5 Mg/Mg and WB-1 Mg/Mg joints represented highest and lowest maximum tensile shear stress respectively; however, the WB-0.5 Mg/Mg joints were slight lower than the AB-0.5 Mg/Mg joints at both low temperature and RM, while it was equivalent to the AB-0.5 Mg/Mg joints at high temperature. The maximum tensile shear stress for all joints decreased significantly from 20°C to 90°C, indicating a stronger influence of high environmental temperature, but it just slightly increased from 20°C to -40°C.

The energy absorption was estimated, which was defined as the area below the load-displacement curve. The evaluated results are plotted in Fig. 8.5(c). The energy absorption of both AB-0.5 Mg/Mg joints and WB-0.5 Mg/Mg joints were much lower than that of the WB-1 Mg/Mg joints. This suggests that the doubled overlap area in the WB-1 Mg/Mg joints greatly enhanced energy absorption. However, it can be seen that at higher temperatures, all values became closer, which implied that the elevated temperature reduced the difference of energy absorption arising from the size of bonding area. The energy absorption of the WB-0.5 Mg/Mg joints was observed lower at the lower temperature compared with the AB-0.5 Mg/Mg joints, while the former was equivalent at a higher temperature and RT.

(a)

(b)

Fig. 8.5 (a) The maximum tensile shear stress, (b) the maximum tensile shear load and (c) the energy absorption versus test temperature for the AB-0.5 Mg/Mg joint, the WB Mg/Mg joint and the WB-1 Mg/Mg joint.

Similarly, Fig. 8.6 shows both the maximum shear load and the energy absorption versus temperature for the WB Mg/steel and RSW Mg/steel joints. It can be seen from Fig. 8.6(a) that the maximum tensile shear load of the WB Mg/steel joints increased slightly with decreasing temperature because the high temperature softened the adhesive upon heating, hence reduced the ability to carry tensile shear loads as discussed earlier and also in [144]. However, the maximum tensile shear load of the RSW Mg/steel joint increased slightly with increasing temperature, which showed the ability of plastic deformation of spot welded joints became better with increasing temperature [31].

Fig. 8.6 (a) The maximum tensile shear load and (b) the energy absorption versus temperature for the WB Mg/steel joint and the RSW Mg/steel joint.

The energy absorption of the WB Mg/steel joints was just slight higher than that of the RSW Mg/steel joints due to high environmental temperature would reduce the impact resistance of the WB Mg/steel joints, however, it was much higher than that of the RSW Mg/steel joints at both lower temperature and RT as shown in Fig. 8.6(b). This reflected the beneficial role of adhesive and its positive influence.

8.3 Tensile Failure Modes

After the tensile shear tests, failure modes of the specimens were evaluated visually. There were several types of bond failure for AB joints [145]: (i) Cohesion failure, where the adhesive fractured, leaving traces of adhesive on both surfaces; (ii) Adhesion failure, where the bond failed at the interface between the adhesive and the adhered material; (iii) Substrate failure, where the bond failed at the substrate; (iv) Mixed-mode failure which was a combination of both cohesion and adhesion failures, sometimes plus substrate failure. It can be seen from Fig. 8.7 that the AB-0.5 Mg/Mg joint failed in the mode of cohesive failure at both lower and room temperatures, while it failed in a mixed failure mode at an elevated temperature.

Fig. 8.7 Fracture surface of the AB-0.5 Mg/Mg joint after tensile shear tests at (a) low temperature, (b) RT, and (c) high temperature.

Figs 8.8 and 8.9 revealed that both WB-0.5 Mg/Mg and WB-1 Mg/Mg joints failed along the adhesive layer in the mode of adhesive failure (Figs. 8.8(a) and 8.9(a)) at a lower temperature and cohesive failure mode (Figs. 8.8(b) and 8.9(b)) at room temperature, respectively. However, at a higher temperature, it is of interest to observe that the tensile fracture surfaces of the WB-0.5 Mg/Mg joints showed a mixed failure mode (Fig. 8.8(c)) on the adhesive layer, while the WB-1 Mg/Mg joints showed an adhesive failure mode (Fig. 8.9(c)) on the adhesive layer with its fracture jumping from one interfacial surface to another irregularly with a coarse pattern. Both WB-0.5 Mg/Mg and WB-1 Mg/Mg joints failed in the nugget with a partial button pull-out failure mode (Figs 8.8 and 8.9).

Fig. 8.8 Fracture surface of the WB-0.5 Mg/Mg joint after tensile shear tests at (a) low temperature, (b) RM, and (c) high temperature.

Fig. 8.9 Fracture surface of the WB Mg/Mg joint after tensile shear tests at (a) low temperature, (b) RT, and (c) high temperature.

Fig. 8.10 shows a failure mode in a WB Mg/steel joint. The WB Mg/steel joint failed in a mixed failure mode (Fig. 8.10(a)) on the adhesive layer with a partial substrate fracture at low temperature. It is seen that the hot-dip galvanized zinc coating layer was peeled up from the surface of HSLA steel specimen and stuck at the Mg side after the sample failure (Fig. 8.10(a)). The EDS spectrum of a line scan analysis in Fig. 8.11 from the fracture surface at the Mg side of the WB Mg/steel joint verified this result. The WB Mg/steel joint failed on the adhesive layer in a mixed failure mode without any substrate fracture at both room and higher temperature as seen in Figs 8.10(b) and 8.10(c). Furthermore, the failure in the nugget of the WB Mg/steel joint was characterized by the interfacial fracture mode.

Fig. 8.10 Fracture surface of the WB Mg/steel joint after tensile shear tests at (a) low temperature, (b) RT, and (c) high temperature.

Fig. 8.11 EDS spectrum of a line scan analysis from the fracture surface at Mg side of the WB Mg/steel joint after a low temperature tensile test in Fig. 7(a) (where purple: iron, red: oxygen, blue: zinc, and green: Mg).

8.4 Summary

- (1) The maximum tensile shear load for the AB-0.5 Mg/Mg, WB-0.5 Mg/Mg, WB-1 Mg/Mg, and WB Mg/steel joints decreased at 90°C, while it increased to a certain extent at -40°C due to the adhesive hardening induced ability to carry more tensile shear loads.
- (2) The energy absorption of the AB-0.5 Mg/Mg and WB-0.5 Mg/Mg joints was lower than that of WB-1 Mg/Mg joint at both low temperature and RT because of the overlap area. However, they were much closed at high temperature, indicating a significant influence of high temperature.
- (3) The maximum tensile shear load of the RSW Mg/steel joint increased slightly with increasing temperature, while the maximum tensile shear load of the WB Mg/steel joints increased slightly with decreasing temperature because of the softening of the added adhesive.
- (4) The energy absorption of the WB Mg/steel joints was higher than that of the RSW Mg/steel joints at both -40°C and 20°C due to the beneficial role of adhesive layer, but it reduced steeply at 90°C with a value close to that of the RSW Mg/steel joint, suggesting a diminished role of the adhesive at elevated temperatures.

CHAPTER 9 CONCLUSIONS

A conclusion of significant results and contributions provided by this thesis work is presented in the following section, which is followed by recommendation for future work.

9.1 Conclusions

- (1) Both DLW DP980 and FLW DP980 joints showed tempered martensite at the outer HAZ which caused softening in the welds, the higher welding speed and less heat input resulted in less martensite tempering or decomposition in the HAZ of FLW joints.
- (2) A superior tensile strength was obtained for the FLW DP980 joints because the higher welding speed produced a much narrower soft zone or a higher hardness in comparison with the DLW DP980 joints or the FLW HSLA joints.
- (3) Although fatigue life of the FLW DP980 joints was not improved at a lower stress amplitude, it was much longer than that of both DLW DP980 and FLW HSLA joints
- (4) The multiple linear welds in fiber laser welding appeared to a larger scatter and increase the probability of dynamic fatigue failure at lower level of stress amplitudes due to an increasing number of weld concavities and soft zones along the welds.
- (5) The added adhesive layer diminished stress concentration and produced much stronger WB Mg/Mg and Mg/steel joints than RSW Mg/steel joints in terms of the maximum tensile shear load and fatigue strength.
- (6) The maximum tensile shear load and energy absorption decreased slightly with increasing crosshead speed (or strain rate) in the RSW Mg/steel joint, while they increased with increasing strain rate in both WB joints of Mg/Mg and Mg/steel. This reflected a beneficial role of weld adhesive in enhancing the energy absorption capacity in vehicle crashworthiness.

(7) Half reduced bonding length in the WB-0.5 Mg/Mg joints resulted in fewer shrinkage pores in nuggets and thus higher maximum tensile shear stress and fatigue resistance. Although it showed a slight lower maximum tensile shear stress and fatigue resistance at higher stress amplitude levels compared with the AB-0.5 joints, the energy absorption for both of them was equivalent.

(8) The maximum tensile shear load for the AB-0.5 Mg/Mg, WB-0.5 Mg/Mg, WB-1 Mg/Mg, and WB Mg/steel joints decreased at 90°C, while it increased to a certain extent at -40°C due to the adhesive hardening induced ability to carry more tensile shear loads.

(9) The energy absorption of the AB-0.5 Mg/Mg, WB-0.5 Mg/Mg, WB-1 Mg/Mg, and WB Mg/steel joints had no distinct performance at 90°C, suggesting a diminished role of the adhesive at elevated temperatures.

9.2 Major Contributions of Present Studies

The research both parts of high power fiber laser welding of DP980 steel joints and weld-bonding of Mg alloys related to similar and dissimilar joints has achieved for the first time the goal of providing beneficial information and data for automotive industry.

(1) Compared with other types of laser welding, the fiber laser welding with a high power density can be used to achieve a much narrower HAZ and FZ, which led to a considerable improvement of the mechanical properties with a joint efficiency in the tensile tests reached as high as 96-97%.

(2) A higher welding speed (16 m/min) during the FLW process generated stronger joints with further higher mechanical properties. This means that both superior quality and high productivity can be achieved using the FLW, in comparison with the DLW process (with a welding speed of 1 m/min).

(3) A key factor in the FLW DP980 steel joints was identified to be the welding concavity which strongly affected the fatigue resistance at the lower stress amplitudes, although it did not influence the joint efficiency in the tensile tests. Once the welding concavity was removed, the fatigue resistance would be expected to improve significantly.

(4) The study related to the WB joints verified that the added adhesive layer effectively diminished stress concentration and produced much stronger WB Mg/Mg and Mg/steel joints than RSW Mg/steel joints. Although the mechanical properties including the maximum tensile shear stress and fatigue resistance of the WB Mg/Mg-0.5 joints were slightly lower than those of AB Mg/Mg-0.5 joints, the ability of energy absorption for both of them was still equivalent. Thus, these results indicated the additional benefits of weld-bonding in the area of welding and joining in the automotive industry.

9.3 Recommendations for Future Work

(1) Since the detrimental influence of concavity was observed in the FLW joints, further studies are needed to eliminate the weld concavity and improve fatigue strength.

(2) The influence of different HSLA steel coating types on such as hot dip galvanizing (HDGI), hot dip galvannealing (GA), and electrogalvanizing (EG) in the FLW process should be evaluated to identify the most suitable coating method for the automotive applications.

(3) In spite of the absence of FZ at the steel side was observed during the joining process of both WB and RSW Mg/steel, the entire welding nugget of Mg/HSLA steel dissimilar welded joints need to be further studied by optimizing of welding parameters, such as welding current, electrode cap size and form, cooling rate, overlap area and adhesive thickness, etc.

(4) Although the reason why the tensile and fatigue strength of WB Mg/Mg-0.5 joints was lower than that of AB Mg/Mg-0.5 joints has been identified, a further study is needed to achieve the equivalent mechanical properties for both types of joints.

(5) The study associated with the temperature dependence of WB joints indicated a diminished role of the currently used adhesive at elevated temperatures, it would be worthwhile to explore alternative types of (or high temperature-resistant) adhesive in the WB process.

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Appendix: List of Publications (over the past two years during my MAsC study)

EDUCATION

Master of Applied Science (August 2012)
Ryerson University, Toronto, ON
Major: Mechanical Engineering

Master of Business Administration (June 2001)
Tsinghua University, Beijing, China

Bachelor of Applied Science (June 1989)
Wuhan Institute of Chemical Technology, Wuhan, China
Major: Mechanical Engineering

AWARDS

Ryerson Graduate Scholarship for academic excellence, 2011-2012
Ryerson Graduate Stipend, 2010-2012
Tsinghua University Excellent Graduate Student and Degree Thesis Award, 2001

LIST OF PUBLICATIONS (over the past two years during my MAsC study)

1. W. Xu, D.L. Chen, L. Liu, H. Mori, Y. Zhou: Microstructure and mechanical properties of weld-bonded and resistance spot welded magnesium-to-steel dissimilar joints, *Materials Science and Engineering*, A 537 (2012) 11-24
2. W. Xu, D. Westerbaan, S.S. Nayak, D.L. Chen, F. Goodwin, E. Biro, Y. Zhou: Microstructure and fatigue performance of single and multiple linear fiber laser welded DP980 dual-phase, *Materials Science and Engineering*, A 553 (2012) 51-58
3. W. Xu, D. Westerbaan, S.S. Nayak, D.L. Chen, F. Goodwin, Y. Zhou: Tensile and fatigue properties of fiber laser welded HSLA and DP980 steel joints, *Materials & Design*, 43 (2013) 373–383.
4. W. Xu, D.L. Chen, L. Liu, H. Mori, Y. Zhou: Microstructure and mechanical properties of weld-bonded and adhesive-bonded magnesium joints (under Dr. D.L. Chen's review)
5. W. Xu, D.L. Chen, L. Liu, H. Mori, Y. Zhou: Tensile properties of weld-bonded and adhesive-bonded joints for high and low temperature (under U of Waterloo Dr. R. Liu's review)

CONFERENCE PAPER

1. W. Xu, D.L. Chen, F. Goodwin, E. Biro, D. Westerbaan, S. Nayak, Y. Zhou: Fatigue resistance of laser welded dual-phase steel joints, Sheet Metal Welding Conference 2012, Detroit, MI, USA

CONFERENCE PRESENTATIONS

1. W. Xu, S.H. Chowdhury, F.A. Mirza, V.K. Patel, and D. Sarkar: Fatigue and reliability of welded magnesium joints, APMA-AUTO21 Annual Conference, Ottawa, May 25-27, 2011.
2. W. Xu, S.H. Chowdhury, F.A. Mirza, V.K. Patel, D. Sarkar, and B.S. Naik: Strength improvement of dissimilar welded Joints using novel process, APMA-AUTO21 Annual Conference, Montreal, Quebec, May 29-31, 2012.