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• Low-Impedance Resistance Welding

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The Many Benefits of Mentoring

As we grow in age and experience, it is important to realize the smallest amount of knowledge we pass on can have the greatest impact on others.

If you refer to the online version of Merriam-Webster’s Dictionary, the word mentor is defined as a trusted counselor or guide. I also feel its meaning can be extended toward acting as a loyal friend and sage advisor. There is no age or status that limits a person to truly being a mentor to someone.

I recently became involved with the Pearland Independent School District’s RISE (Reach, Inspire, Support, Empower) Mentoring program. I am currently mentoring a student at Robert Turner High School in Pearland, Tex. I was matched with him because he is in his third year in the school’s welding program.

It has been very interesting to see things through the eyes of a 17-year-old student. He wants to enter the welding industry after graduation. In our short time together, I believe he has given me as much insight from his generation as I have to him. It just shows that sharing information can be a great tool in our own development no matter what the age group.

In the vast diversity of generations of men and women in the welding industry, it is important to realize just how difficult it may be to communicate with one another. Although a lot of the welding family is experienced in a different generation, that knowledge is still very valuable. Statistics show, for the first time, there are five generations in the workplace. Becoming a mentor does have some challenge when it comes to communication, but that does not mean we can’t try.

In the past year, Vice President Thomas Lienert, along with his committee members, started a program for mentoring American Welding Society (AWS) board members. It began for incomers (district directors, directors at large, the vice president elect), but after realizing how important it was to ensure current members could be mentoring prospective replacements from their district, this program has extended to everyone on the board.

Incoming board members will attend the spring and fall meetings for their introduction. With the help of AWS Corporate Director, Member Services, Rhenda Kenny, they have already rewritten the manual for district directors. All of this shows just how much the sharing of knowledge and ideas can impact so many people.

These are the very reasons I am going to ask everyone I come in contact with this year to consider being a mentor to just one person, to stay in contact with this person, and be a sage advisor who can turn into a loyal friend.

As we all know, you have to want to be a mentor to be a good mentor. I believe once you get started and can see how your information can impact your mentee, you will want to continue being their mentor.

When I have the privilege of writing the Welding Journal Editorial in the December 2017 magazine, I will have the chance to share with everyone just how our mentoring plan was received by all this year.

By being a mentor to someone new in the welding family, this idea could spread all over the world if people would just take the time to try.

Similar to what past President and Chairman of the AWS Foundation Bill Rice stated, “We are looking for welders, not just new AWS members.” That theme of introducing new people to welding is what impacts all of us. No matter the extent of one’s education experience, we all can qualify to be a mentor to someone in the welding industry.

So, in the end, it’s still all about our best asset — people. If we can broaden our welding family and help strengthen our trades in this world, we all will benefit from everyone’s efforts.

“By being a mentor to someone new in the welding family, this idea could spread all over the world if people would just take the time to try.”
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Wall Colmonoy Awarded Contract for Overhauling F-16 Heat Exchangers

Wall Colmonoy’s Aerobraze Engineered Technologies, Oklahoma City, Okla., has won a competition by the United States Air Force (USAF) for a multiyear, multimillion dollar contract to overhaul the F-16 primary and secondary heat exchangers. The F-16 has dominated the skies for more than 30 years as a multirole and air fighter in service in more than 25 countries.

“As a major manufacturer and overhaller of heat exchangers for the USAF for over 33 years, it is with great pride that we continue to provide support to the United States Air Force. This new award demonstrates our ongoing commitment to increasing efficiency and cost savings for our customers, through extending the life of critical components for the aerospace industry,” said Brian Martin, general manager for Aerobraze Oklahoma City.

Florida State College at Jacksonville Receives Nearly $2 Million Grant for Workforce Development, including Welding Training

Florida State College at Jacksonville has been awarded America’s Promise Grant worth $1,804,656 from the U.S. Department of Labor to expand job training partnerships. The funds will be used to support the School of Science, Technology, Engineering, and Mathematics through advanced manufacturing training opportunities, specifically in welding and mechatronics technologies.

The grant, inspired by President Obama’s America’s College Promise Grant plan to allow two free years of community college for responsible students, is designed to accelerate the development and expansion of workforce partnerships that provide skilled workers in specific, in-demand sectors.

The award allows for a project that aims to develop and implement an accelerated 10-week core fundamentals boot camp on building essential manufacturing workforce skills as identified by employer partners, a Core+ upskilling training component that can be customized to employer needs, an America’s Promise Manufacturing Open Lab, and skills attainment through Work & Learn paid internships.

Over the four-year project period, the college will serve a minimum of 250 participants from many populations.

Credentials to be awarded include the following: American Welding Society Basic Welder, Occupational Safety and Health Administration 10 and 30, Manufacturing Skill Standards Council-Certified Production Technician, National Institute for Metalworking Skills Level 1, and Autodesk Certified User-AutoCAD and Professional-AutoCAD.

Huntington Ingalls Industries Launches Arleigh Burke-Class Destroyer Paul Ignatius

Huntington Ingalls Industries’ Ingalls Shipbuilding division has launched Paul Ignatius (DDG 117), the company’s 31st Arleigh Burke-class (DDG 51) guided missile destroyer.

Named in honor of the Secretary of the Navy from 1967 to 1969 and the Assistant Secretary of Defense during President Lyndon B. Johnson’s administration, Paul Ignatius was translated via Ingalls’ railcar system to a floating dry dock. Once on, the dry dock was moved away from the pier, and it was ballasted to float the ship.

“The DDG 51 program provides our U.S. Navy customer and our nation a series of highly advanced and capable warships,” said Ingalls Shipbuilding President Brian Cuccias.

“For 30 years, our talented shipbuilders have been building these much-needed, quality destroyers. Launching DDG 117 is an important milestone in the life of the ship, which will continue building toward fleet readiness in 2018.”

AT&F’s New Robotic Welding System Handles Complex Design Features

AT&F Cleveland, Ohio, a provider of custom and high-volume steel fabrication, has added a large robotic welding system to its line of capabilities. The multiaxis programmable welding arm can perform consistent welds at high speeds and control complex welding situations.

In addition, the system is capable of welding lengths up to 40-ft long and 20-ft high. Six axes on the arm allow for a far range of motion, while two sets of 30,000-lb turning rolls and a 120,000-lb turntable position the heavy workpiece. It has a maximum pass speed of 30 in./min.

Installed in the robot’s arm is an optical laser system that determines the position and orientation of a workpiece. This autonomous correction feature produces welds as long as 40 ft.
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Axalta Announces Plan to Build a Manufacturing Facility in Nanjing, China

On Nov. 11, Axalta Coating Systems, an international supplier of liquid and powder coatings, publicized its plan to build a manufacturing plant for high-performance cars, commercial vehicles, and industrial coatings in Nanjing, China. The facility will be built in the Nanjing Chemical Industrial Park and is expected to take up more than 170,000 square meters of land.

It will be used to manufacture an expansive collection of the company’s products, such as primers, basecoats, clearcoats, and resin intermediates, as well as the AquaEC™ line of electrocoats and the Voltatex line of energy solution coatings. It will also function as a logistics and distribution center for the company’s refinish and industrial products, specifically serving customers in southern and central China.

The Nanjing facility is expected to be fully operational by the end of 2020. Once completed, it will be the company’s third-largest operations facility.

“We look forward to continued opportunities to grow in China,” said Luke Lu, Axalta vice president and president of Axalta in greater China. “We anticipate allocating between $100 and $150 million over the next three to five years toward the Nanjing project.”

Company Celebrates Grand Opening of Demo and Technology Center in Mexico

Kurtz Ersa Mexico, S. A. DE C. V., a subsidiary of Kurtz Ersa North America focusing on sales and service of soldering systems, celebrated the grand opening of its demo and technology center in Guadalajara, Mexico, on Nov. 17.

“With the new subsidiary in Guadalajara, we are in a position to act more swiftly on the Mexican market — with competent service from our local, Spanish-speaking colleagues. The central Ersa Mexico warehouse will also be located here in Guadalajara — as will the Ersa Demo Center and the general distribution,” said Albrecht Beck, Kurtz Ersa, North America managing director.

In attendance were global and local partners, as well as local customers. The event was marked with a ribbon-cutting ceremony, tours of the new center, and live demonstrations of the company’s soldering systems, including the Hotflow 3/20, Versaflow 3/45, and HR 600/2.

ASTM International Kicks Off Ambassador Program in Lima, Peru

ASTM International, a global standards organization, delivered its International Ambassadors program geared toward connecting top experts with engineers in Latin America who want to learn and implement ASTM standards.
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AWS Names 2016 Excellence in Welding Awards Winners

Posing for a group shot are the 2016 Excellence in Welding Awards recipients (from left) Robert (Bob) L. Doan (small business), James Daniels (veteran), Jeff Imrie (educator), Frank Witsil (media), Michael Krupnicki (educational facility), Uwe Aschemeier (individual), and Dick Alley (local section). (Not pictured is the large business honoree.)

The American Welding Society (AWS) and WEMCO, an association of welding manufacturers, honored the 2016 Excellence in Welding Awards winners on November 16 at FABTECH in Las Vegas, Nev.

Individual — Uwe Aschemeier. Aschemeier started his career in 1977 as a mechanical fitter in Germany before earning a bachelor’s degree in mechanical engineering and then becoming a Certified Welding Engineer through the German Welding Society. In 1995, he immigrated to the United States. Currently, he is the senior welding engineer for Subsea Global Solutions, a diving company specializing in underwater ship repair. He is chair of the AWS Cincinnati Section, the AWS District 7 director, and has been a member of the AWS board since 2013.

Educator — Jeff Imrie. Imrie has been in the welding workforce for 40 years. He has served as a general welder, multicoded welder, welding foreman/instructor, and college department head. He also works as an educator at Dundee and Angus College in Scotland and is known to partner with each student to create a personalized development plan. He works with companies to place his students into jobs as well.

Educational Facility — Rochester Arc + Flame Center. This facility is the brainchild of Michael Krupnicki, president of Mahany Welding Supply. Between 2002 and 2011, Mahany trained more than 3000 welding students. Realizing there was a need for more courses, Krupnicki opened the center in 2012. It has 85% employment placement within three months of graduation.

Veteran — James Daniels. A Navy veteran with 23 years of experience, Daniels is an AWS Certified Welding Educator and Certified Welding Inspector. He has worked as an instructor at the Tennessee College of Applied Technology – Newbern and grown its welding program from about 10 full-time welding students in one class to more than 40 full-time welding students over three programs.

Large Business — Jeffboat. This manufacturing division of American Commercial Barge Line got its start nearly 80 years ago. Jeffboat’s operation spans 68 acres on the Ohio River. The company provides welding equipment to local schools and creates a pipeline of students who have the chance to become employees.

Small Business — Flange Wizard. Founder and owner Robert (Bob) L. Doan started the company out of his garage in 1981 with one patented tool, the “Flange Wizard.” Today, 35 years and four locations later, the business has grown to 18 employees and a line of more than 50 welding tools and accessories. It has made instructional videos for local colleges to use as part of their welding curriculum as well.

Local Section — Indiana Section. This AWS Section is active in lifting up the welding industry not only in Indiana but through the entire Midwest. They helped develop and administer the Midwest Team Welding Competition; before the contest, members transport welding equipment to the competition and set it up, then donate time to supervising and judging. The Section is also involved with SkillsUSA.

Media Award — Frank Witsil. Witsil won for his article in the Detroit Free Press, titled Women Who Weld: Male-Dominate Skill Aims for Parity, about the nonprofit group called Women Who Weld. He is a Web editor with the newspaper and writes mostly about business. “It’s a privilege to be part of the story rather than just writing about it,” Witsil said. “I’d like to thank the women I wrote about for sharing their stories.”

North Central Michigan College, Industrial Arts Institute Create Welding Program

Starting this month, North Central Michigan College, Petoskey, Mich., and the Industrial Arts Institute, Inc., on- away, Mich., have a new associate degree welding program that can be finished in four semesters with full-time attendance. General education and a portion of the technical skills coursework will be completed through the college. The remaining technical skills coursework will be done through an agreement with the Industrial Arts Institute.

“A trained workforce is essential for manufacturing growth in Northern Michigan,” said Dr. Peter Olson, vice president of academic affairs and student success at North Central. “We began working with our area employers in 2014 with the creation of our CNC (computer numerical controls) program through our mobile fabrication lab. That program continues to grow and expand, and the welding program will add another training opportunity for individuals in our community to find new careers in our area.”

Weld Like a Pro Wins Silver Medal at the International Automotive Media Competition

Weld Like a Pro: Beginning to Advanced Techniques written by Jerry Uttrachi, a past president of the American Welding Society and president of WA Technology, Florence, S.C., has won a silver medal at the International Automotive Media Competition in the book writing technical how-to category. Published in 2015 by CarTech Books, it is similar to Uttrachi’s first book but includes nonautomotive applications, such as welding items for use in his home including modify-
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Glendale Manufacturing Offers Welding and Steel Fabrication Advances in U-Haul System

Glendale Manufacturing Co., Glendale, Ariz., the manufac-
turing arm of U-Haul International, has opened a new
114,000-sq-ft facility. It will expand its welding, steel fabri-
cation, powder coating, and other operations previously ful-
filled at AMFORM (American Fabricators of Raw Materials).
AMERCO, U-Haul’s parent company, acquired the 28-acre
site in 2015.

“There are so many reasons that justify having this facili-
ty, but the main reason is for our team members,” said
Cedric Moore, plant president. “This plant will greatly improve efficiency and employee satisfaction by creating a better working environment, which in turn will improve customer satisfaction with our products and services.”

The site has twice the interior space and nearly six times the acreage of AMFORM. Two fiber lasers are at the center of a computerized laser-cutting and tiered pallet storage system. Six robotic units will handle much of the welding workload. Manual welding areas will service delicate assignments. There’s also an automated line powder coating system.

The plant will fabricate various truck and trailer components, dollies, awnings, key-drop boxes, carriages/housings, tools, and U-Haul display signs. The manufacturer is budgeted to employ 199 team members.

**NUCOR Donates Nearly 90,000 Lb of Steel to Support Different Welding Programs**

NUCOR Steel has recently made two large donations to welding schools. Its Darlington, S.C., location has given 24,146 lb of steel to the welding department at Florence-Darlington Technical College, Florence, S.C. The college usually receives two steel gifts a year from the company; this will supply the department through the fall semester and into the early weeks of the 2017 spring semester.

“It would be difficult to operate without NUCOR’s support,” said Welding Director Jamie King, noting donations allow students to get training on a variety of steels.

In addition, the company’s West Seattle, Wash., division has contributed about 65,000 lb of steel to South Seattle College’s welding fabrication technology program. This will help students hone their skills and prepare for welding careers for several quarters to come.

Welding Instructor Doug Rupik noted the steel bars will be used for classwork, unique building projects, welding booth improvements, and “to help us get students prepared for the Washington Association of Building Officials welding certification test.”

— continued on page 95
Billion Dollar Contract Puts Arconic Sheet and Plate on Every Airbus Platform

Arconic’s $1 billion multiyear contract with Airbus gives both its aluminum sheet and plate a place on the aircraft manufacturer’s highest volume programs.

This agreement is the first to include material from Arconic’s new thick plate stretcher in Davenport, Iowa. The provider’s most significant contribution in the contract is with the A320 aircraft family.

The multiyear agreement starts this month. It makes Arconic sole supplier to Airbus for specific applications, including some wing, fuselage, and structural components.

In addition to Arconic’s proprietary alloys, the planes will include the provider’s plate products on every platform, used in key applications such as wing ribs, fuselage frames, and other structural parts.

Arconic’s thick plate stretcher will come online in 2017. The machine enables producing large, high-strength monolithic wing ribs. Located at the Davenport Works facility, it enhances the performance of thick aluminum and aluminum-lithium plate. The stretching process reduces stress introduced into the plate during manufacturing, resulting in more easy processing.

Global Welding Technologies Launches International Consulting Business

Global Welding Technologies, LLC., St. Louis, Mo., has started an international consulting business. The firm’s focus will be to partner with welding products manufacturers and distributors to develop strategies that drive transformational growth. It will help create plans for improved performance by reviving brands, reversing negative sales trends, and bringing products to the market.

“At our core, we fundamentally understand welding processes, applications, the end segment customer, market trends, and distribution channels,” said President David Wilton. For more information, visit gwt-consulting.com.

GE and Local Motors Form New Business Model for Agile Manufacturing

GE and Local Motors have introduced Fuse™, a new manufacturing approach that accelerates product and technology development by combining open innovation with small batch manufacturing.

The model’s digital community, fuse.ge.com, brings together entrepreneurs, scientists, coders, engineers, and makers around the world to solve product development challenges.

Physical operations will start in microfactories to combine GE teams, customers, entrepreneurs, student groups, and more. These operations will include rapid prototyping, small-batch manufacturing, and modular experimentation.

The first microfactory was launched last month and focused on nondestructive examination answers within medical equipment imaging and product inspection.

Fuse’s creation also serves to launch a new division within Local Motors, called Forth, to provide the platform and services that make co-creation achievable.

Trescal Acquires Metrosul

Trescal, Paris, France, an international specialist in calibration services, has acquired Metrosul, a calibration services provider in the Porto Alegre capital of Rio Grande do Sul, Brazil. This transaction has been completed with the support of Ardian, its majority shareholder. Metrosul founders Nilo Ardais and Claudio Herman will remain in their current positions to continue leading business growth.

Stevens Institute of Technology Dedicates ABS Engineering Center

A dedication ceremony has recently taken place for the ABS Engineering Center at the Stevens Institute of Technology, Hoboken, N.J.

ABS Chairman, President, and CEO Christopher J. Wiernicki addressed attendees at the event. He not only encouraged students to use the facilities to work toward solving industry challenges but also noted they provide an environment for learning engineering disciplines.

The main floor and atrium will contain five laboratory spaces for systems integration, fluids, structural and building materials, robust field autonomy, and naval engineering. A second floor will have 13 faculty offices and workstations for up to 16 graduate students. The third floor will include a 28-seat space for seminars and project presentations.
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- Opposing Foreign Steel Dumping in Canada
- What's a Public-Private Partnership, and Why Is It Important to Me?
- How to Work Effectively Across the Generations: From Baby Boomers to Millennials and Beyond
- So, What Should I Measure? Metrics That Matter to Local Unions and Contractors
- Weld Cloud
- IW Safety Director Course
- Skidmore – Bolting
- Putting the Pedal to the METAL – Metal Building’s Growth and Future
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Q: We laser beam weld various medical device components. Some are made of stainless steel alloys and some are made of titanium alloys. We use 5-9 argon shielding gas for all welds. Another company uses compressed air as the shielding gas. They laser weld automotive components made of carbon steel. Is there any reason for us to use argon, which is more expensive than air? More fundamentally, how do we select the correct shielding gas for laser welding?

A: This is a good question, as it addresses both technical and economical issues related to shielding gases used in laser welding. Let’s take an overview of the subject and provide guidelines and the answers to your questions.

**Overview.** Consider the following key pieces of information about laser welding, which were covered in the September and November 2016 Laser Welding Q&As, as well as some general information about shielding gases and references.

1. Focused laser beams produce very high power densities on the surface of the material.
2. The rates of heating the metal and the cooling rate of the metal are extremely high.
3. Surface tension in the molten pool of laser welds is very high. The high surface tension is mostly the result of the temperature gradient across the molten pool. When the metal solidifies, the surface tension becomes residual stress in the material and may cause failure of the weld.
4. Quality of the argon gas used for shielding gases is critical, since one of the contaminant gases is oxygen. Oxygen is known to affect the surface tension of the molten pool. It is recommended that you use the ‘5-9’ argon gas, which is an ultrahigh purity grade where the purity of the gas is 99.999%. This quality of gas is only available in compressed gas cylinders (not in liquefied gas dewers).
5. The presence of metal vapors plus gases that can ionize, high power density, and high average power at the surface of the metal create intense plasma that may affect penetration and other weld qualities. As laser powers of 8 kW and higher, and at power densities values of E+09 W/in.\(^2\) (E+08 W/cm\(^2\)) and higher, the plasma may be controlled with the use of helium shielding gas.

6. If compressed air is to be used as a shielding gas, it must be filtered for moisture, solid particles, and oil vapors. Consider refrigerating the air to freeze all residual moisture. Please read the Safety section below.

7. AWS C5.10/C5.10M:2003, Paragraph 4.6, “Recommended practices for shielding gases for welding and cutting” offers a basic overview of the subject.

8. AWS C7.2M:2010, Paragraph 7.5.6, “Recommended practices for laser beam welding” is also general. The AWS C7C Committee is currently revising this AWS standard.

**Safety**

In some applications, it is possible to use compressed air as a shielding gas. Remember safety when selecting the shielding gas especially with reactive (exothermic) metals. These metals readily absorb ambient oxygen at elevated temperatures, then oxidize and release a large amount of heat. The amount of heat released is sufficient to heat additional material that absorbs oxygen and the process perpetuates and goes out of control, i.e., the self-burning process is hard to control.

**Function of Shielding Gases**

In laser welding, experience indicates the main functions of the shielding gas, as are follows, in order of importance:

1. Protecting the focusing optics.
2. Protecting the surface of the molten metal.
   This means that one type of gas may be used to protect the optics and another type of gas may be used to shield the molten metal.

**Delivery of Shielding Gases**

The key parameters of the gas delivery systems are flow rate and flow velocity, as measured at the surface of the metal being welded. The gas flow should be laminar. Turbulent gas flow tends to mix some of the ambient oxygen into the shielding gas. The diameter of the gas jet is typically 0.250 in. (6.0 mm). The distance between the end of the gas delivery tube and the material being welded is typically 0.250 in. (6.0 mm). Diffusers like those used in arc welding are not recommended as they cause too much turbulence. Typical flow rates are in the range of 20–50 ft\(^3\)/h (9.5–23.8 L/m), and flow velocities in the range of 4.5–11 mph (2–5 m/s). The gas delivery may be coaxial with the focusing optics or could be auxiliary, arranged to produce gas flow in the direction of welding.

**Surface and Bulk Effects of Shielding Gases on Weld Quality**

To minimize surface tension and the residual stresses in the weldments, use 5-9 argon or helium. (see No. 4 in overview.)

The plasma should be suppressed so weld penetration is more consistent. Typically, when the average laser power is 8000 W and higher, helium should be used. (see No. 5 in overview.)

**Bulk Effects on the Material**

For nonreactive metals, it has been shown that it is the cooling rate of the weld and not the shielding gas that determines the metallurgical structure of the solidified metal. Further, the shielding gas does not affect weld strength. An article published by EWI (Ref. 1) compared the strength of laser welded bar stock samples that used air, argon, and helium. The results were essentially the same and independent of the shielding gases.

Absorption of gases in the molten metal is minimal because of the very fast cooling rate of the molten metal. Exceptions are the reactive metals such as titanium and its alloys, and magnesium and its alloys.

Weld penetration, however, is a function of the type of shielding gas used. Deepest weld penetration is achieved by using helium. In Class II metals, argon, nitrogen, and air produce approximately the same weld penetration.

Weld penetration is also affected by the direction of the shielding gas flow.
There is a vectorial relationship between the direction of welding and the direction of gas flow. This means that during laser welding, the direction of the flow of the shielding gas should not change.

**Now to Answer Your Questions**

Weld penetration in medical devices is typically in the range of 0.004–0.040 in. (0.1–1.0 mm). Further, the average power of the laser is much under 8000 W. Often, pulsed laser power is used; therefore, plasma formation is typically not an issue. However, the residual stresses and the aesthetics of the welds are critical. For these reasons, it is recommended to use helium or the less expensive 5-9 argon gas. Apply the shielding gas on both the top and underside of the weldments. Note that the surface tension and the residual stresses are also affected by other process parameters.

Weld penetration in automotive components is typically in the range of 0.040–0.120 in. (1.0–3.0 mm). The average power of the laser is typically in the range of 4000 to 6000 W. Plasma formation is very much an issue. Residual stresses and aesthetics may not be an issue. The first choice of shielding gas would be helium — and for many years it was; however, when using compressed air the plasma is manageable, and much less expensive. The slightly reduced weld penetration with air is easy to fix: use more laser power or slow down the welding speed. The automotive industry usually opts for the higher laser power.

Reference

Q: For many years, we joined carbon steel and low-alloy steel to stainless steels like 304L and 316L using ER309LSi filler metal. Part of our quality assurance program was to check ferrite on the weld metal with a magnetic instrument. Now some customers are asking us to use ERNiCr-3 filler metal for some of those joints. They also want us to measure ferrite on those welds for a procedure qualification record (PQR). I wonder if this is appropriate.

A: I have been asked this question on a number of occasions. To begin, a small amount of ferrite (approximately 3 to 4 ferrite number or FN, minimum) is very helpful in assuring that solidification occurred in the primary ferrite solidification mode, thereby providing resistance to solidification cracking in nominally austenitic stainless steel welds. This is a situation specific to nominally austenitic stainless steel weld metals. The ferrite can be found by metallographic cross section, which is a destructive test, or nondestructively by using an instrument calibrated to AWS A4.2, Standard Procedures for Calibrating Magnetic Instruments to Measure the Delta Ferrite Content of Austenitic and Duplex Ferritic-Austenitic Stainless Steel Weld Metal. The nondestructive method is more popular. The ferrite phase is ferromagnetic, so it responds to a magnet or an induced magnetic field. The magnetic response can be measured to determine the amount of ferrite present.

In a weld of carbon steel or low-alloy steel to a stainless steel such as 304L or 316L made with ER309LSi filler metal, ferrite content will nominally exceed 5 FN, and the weld metal will be highly resistant to solidification cracking. The bulk of the weld metal will be austenite, which is not ferromagnetic. However, the carbon steel or low-alloy steel is ferromagnetic, so you must make your measurement point on the weld centerline, not near the carbon steel or low-alloy steel.

Ferrite and carbon steel or low-alloy steel are not the only ferromagnetic phases that can be found in welds. Anything else that is ferromagnetic will give a false high ferrite number. In the dissimilar metal joints you mentioned, enough dilution is possible in some circumstances (for example, a submerged arc weld made with ER308L filler metal) to allow the weld metal to transform to martensite.

Most martensite in iron-based alloys is ferromagnetic, although a nonmagnetic epsilon martensite can be found in some alloys such as austenitic manganese steel weld metals.

Ferromagnetism arises when the magnetic domains in a metal align in the presence of a magnetic field. When the temperature is increased, a temperature can be reached where the ordered magnetic domains become disordered. This temperature, called the Curie temperature, after Pierre Curie who discovered the effect in the nineteenth century, is the highest temperature at which the metal is ferromagnetic. Pure iron, carbon steels, and low-alloy steels all exist as body-centered cubic (BCC) crystals at ambient temperatures, as does the ferrite in nominally austenitic stainless steel weld metals. At a temperature of about 1670°F (910°C), pure iron transforms from BCC crystals to face-centered cubic (FCC) crystals, austenite. But 1670°F and the phase transformation have nothing to do with the Curie temperature of pure iron, which is about 1420°F (770°C). So the BCC crystal structure is not always ferromagnetic.

Austenite in stainless steels is FCC and nonmagnetic, but commercially pure nickel (Nickel 200, UNS N02200) is also FCC, and it is ferromagnetic (Curie temperature about 680°F or 360°C). Cobalt, the third relatively common metal that is ferromagnetic, has a hexagonal, close-packed (HCP) crystal structure, and a very high Curie temperature of about 2039°F (1115°C).

Some nickel-based alloys are ferromagnetic at room temperature, and some are not. ERNiCr-3 weld metal is not ferromagnetic, so any attempt to measure ferrite with a magnetic instrument will essentially yield zero as a test result, and metallography will not find any ferrite. On the other hand, alloys containing anywhere from 36 to 81% nickel, balance iron, are ferromagnetic. One of these, commonly called Invar (UNS K93603), nominally 36% Ni, balance Fe, FCC crystal structure, about 535°F (279°C) Curie temperature, has an outstandingly low coefficient of thermal expansion, with matching filler metal for which I could find no specification number. An unusual alloy that sometimes displays ferromagnetism is Monel® Alloy 400 (UNS N04400) (Ref. 1), which consists of approximately 65% Ni, 30% Cu, and on the order of 1% Fe and 1% Mn. The alloy’s crystal structure is FCC, and its Curie temperature ranges from 70° to 120°F (21° to 49°C), so that some heats of this alloy are ferromagnetic at ambient temperature and some are not. A heat that is not ferromagnetic will become ferromagnetic if cooled below room temperature.

The point of all of the above information is to indicate that magnetic ferrite measurements on anything other than nominally austenitic stainless steel weld metal or duplex ferritic-austenitic stainless steel weld metal, as specified in AWS A4.2, at best will be useless and at worst could be very misleading.

In particular, measuring ferrite when joining 304L or 316L stainless steel to carbon steel or low-alloy steel using ERNiCr-3 filler metal, or any other nickel-based alloy filler metal, will yield no information that is of any use in assessing whether or not solidification cracking in the weld metal is likely. I advise against including such measurements in a PQR or any quality assurance program covering this situation.

Reference
1. See the online bulletin for Monel Alloy 400 of Special Metals Corp., specialmetals.com.
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Q: Our company splice welds steel bar stock of various identical diameters using the flash welding process. The bar stock is normally spliced end to end; however, a customer has asked us for a particular application for which we would try to weld the bar stock at an angle to each other. We have tried various settings on the welding machine, including increasing the current. Unfortunately, we have excess molten metal protrude at the apex of the joint and, consequently, this area becomes considerably weakened. Is there some way to overcome this?

A: You do not mention the degree of angle of the bar stock joint, but your statement that you have molten material protruding from the apex of the weld leads me to believe you have surpassed the permissible angular intersection of the round bar stock. For flash (resistance) welding, it can be no greater than 15 deg off the centerline of the offsetting movement. The reason for the molten slag protruding from the apex area is due to not having sufficient backup of base material (and possibly weld die) adjacent to the weld interface.

In the normal splicing of the bar stock end to end, the flash produces heat in the joint, creating a heat-affected area wide enough on the two ends of the bar stock to upset (forge) the two ends together. In the angle application, the apex of the weld does not have sufficient material behind the heat-affected area, resulting in poor upset by allowing the heat-affected area to be expelled outward away from the joint. The increase in current you needed was due to the increase from a circular joint to an elliptical joint, which increased the area of the joint to be welded.

Some experimental weld die design to incorporate a backup of the apex area of the joint has proven to help but will not always substitute for the lack of sufficient base metal behind the heat-affected area. The results would depend on your stress values and the actual application. As a rule of thumb, the included angle of a joint of round bar stock must be at least a minimum of 150 deg to get a satisfactory upset.

LARRY MOSS is president of Automation International, Inc., Danville, Ill., and has been in the resistance welding field for 45 years. He is a past president of the Resistance Welding Manufacturing Association, which is now the Resistance Welding Manufacturing Alliance (RWMA), a standing committee of the American Welding Society. He was the 2009 recipient of the Elihu Thomson Resistance Welding Award from the American Welding Society, and has been a weld school instructor for RWMA for the past ten years. Send your comments and questions to Larry Moss c/o Welding Journal, 8669 NW 36 St., #130, Miami, FL 33166, or via e-mail at lmoss@automation-intl.com.
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Global Market Study on Welding Equipment: By Welding Technology, Arc Welding Segment is Estimated to Account for 47.6% Value Share by 2016 End analyzes the welding equipment market at both the global and regional levels over an eight-year forecast period (2016–2024). The report discusses the market dynamics expected to influence the current environment and future status of the global welding equipment market. The report also provides updates on trends, drivers, restraints, value forecasts, and opportunities for manufacturers.

Standard Addresses Directed Energy Deposition

F3187, Guide for Directed Energy Deposition of Metals, includes directed energy deposition (DED), a metal additive manufacturing process. DED is utilized for repair, rapid prototyping, and low-volume part fabrication. The standard was developed by the ASTM International Committee F42 on Additive Manufacturing Technologies.

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Report Forecasts Resistance Welding Machine Market Growth

Resistance Welding Machine Market Growth Forecast Analysis by Manufacture...
turers, Regions, Type and Application to 2021 provides a competitive analysis of the resistance welding machine market. Some of the manufacturers covered are ARO Technologies, NIMAK, Fronius International, T. J. Snow, Panasonic Welding Systems, and Taylor-Winfield. Spanning 115 pages, the report also focuses on the resistance welding machine markets in North America, Europe, Asia-Pacific, Latin America, the Middle East, and Africa.

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FABTECH 2016 Comes Up a Winner

BY ANDREW CULLISON, KRISTIN CAMPBELL, CARLOS GUZMAN, CINDY WEIHL, AND MARY RUTH JOHNSEN

This year’s show offered plenty of new welding products, as well as education and networking opportunities.

No matter how attendees decided to spend their time at this year’s FABTECH, they were betting on a sure thing since so many activities were jammed into the three-day event. The 31,000 attendees from more than 120 countries could choose from a robust professional program, expert-led panel discussions, or myriad product demonstrations on the show floor.

This year’s show was held November 16–18 at the Las Vegas Convention Center. At more than 575,000 net sq ft of exhibit space, this was the largest-ever FABTECH held in Las Vegas. The exhibition is sponsored by the American Welding Society (AWS); SME; Fabricators & Manufacturers Association, International; Precision Metalforming Association; and Chemical Coaters Association International. It is the largest annual event dedicated to metalforming, fabricating, welding, and finishing in North America.

No One is Born a Champion

Sugar Ray Leonard delivered an inspiring opening keynote speech titled The Power to Win to a packed FABTECH audience — Fig. 1. Leonard, the author of The Big Fight: My Life In and Out of the Ring, spoke to the standing-room-only crowd in an enthusiastic manner and on a personal level. He said he wasn’t there to speak of his physicality but to talk about the power of the mind. This six-time champ in five-different weight classes spoke of his philosophy. “Your dreams are win-

Fig. 1 — Sugar Ray Leonard entertained a packed house during his keynote address The Power to Win.
dows to the future,” he said. “They are the GPS to your success. No one is born a champion. Success comes from your desire to do what you want. It is in your brain.”

His lessons on life were brought home with dramatic videos of his famous fights. He talked about how after retiring and not having a fight in five years, he announced he was going to fight “Marvelous” Marvin Hagler who was in his prime. Everyone thought he could never win, but he was determined to succeed and his path to success was to put in the extra roadwork. Instead of sparring three rounds with a minute rest in between, he went five rounds with 30 seconds between. His workouts were designed to build up his stamina to almost the breaking point. The sting in his legs was the signal he was reaching his goal. “My determination allowed me to do what many thought impossible,” he said. His upset victory is still one of the legends of boxing history.

Roadwork and determination are reoccurring themes in Leonard’s life. “To turn life around, you have to do the roadwork. It makes you dig deep and builds up the reserve you need to activate it when it comes time.”

Self-esteem is another character trait that Leonard sees separating winners from losers, and he noted roadwork has a tremendous impact on self-esteem. He told of an episode in his training camp five days before his fight with Hagler. His sparring partner almost knocked him out. He spoke of how quiet his training staff became after that incident. “I knew what they were thinking, ‘If a sparring partner could almost take him down, what was a puncher like Hagler going to do to him?’ I took this serious negative and turned it into a positive.”

Another lesson in life is to “stay in control and don’t lose your composure.” He failed to follow that maxim in his first fight with Roberto Duran. Duran’s prefight behavior affected Leonard’s composure. “He not only insulted me, but he insulted my wife.” Leonard ended up losing after fighting a very uncharacteristic fight for him. “When we aren’t satisfied with what we get in life or business, we have to get up after a knockdown.” He set up a rematch with Duran six months later, which is very soon in the fight game. This time he was composed and on his game. Duran was so frustrated he ended up quitting in the middle of the eighth round supposedly uttering his famous “no mas,” which again is a part of fighting lore.

Risk taking is another of Leonard’s mantras. His biggest risk in boxing was to move up two weight classes to take on Donny Lalonde, the champion in that class. Lalonde was heavier than Leonard and a stiff puncher. Leonard won because “of determination, roadwork, belief in myself, and staying in control.”

Whether it is in your personal matters or business ventures, all life is a fight. Leonard’s final advice for success was “dream your dreams, but have a blueprint to attain those dreams.”

AWS Annual Business Meeting

AWS President David McQuaid called the 97th AWS Annual Meeting to order on Wednesday, November 16. McQuaid detailed the highlights of his year as president. He noted the Society’s accomplishments throughout the year and acknowledged the many people who helped AWS achieve that success and who made his year so fulfilling. McQuaid noted the importance of the Society’s “human capital,” saying “The AWS volunteers, staff, and members make all the difference to the organization.”

That human capital, especially with relation to mentoring, will be the theme for incoming President John Bray’s year. “Never in my wildest dreams did I ever imagine standing here,” he said. Bray then went on to describe the many mentors he has had throughout his life, beginning with his parents and siblings and continuing on to those related to his business and AWS careers.

He challenged the audience to become mentors. “We have a thousand years of experience in this room,” Bray remarked. He mentioned Pearland Independent School District’s RISE (Reach, Inspire, Support, Empower) Mentoring program, through which he mentors a 17-year-old welding student. “I’ve learned more from him than he’s learned from me,” he said.

Bray’s challenge to the audience was to “mentor at least one person and we’ll grow our welding family.”

Following the annual meeting, Paul T. Vianco, an AWS Fellow and Distinguished Member of the Technical Staff at Sandia National Laboratories, Albuquerque, N. Mex., delivered the 2016 Comfort A. Adams Lecture — Fig. 2. Vianco’s talk, titled Understanding the Reliability of Solder Joints Used in Advanced Structural and Electronics Applications, is believed to be the first Adams Lecture on solder joints. The lecture will be published in two parts in upcoming issues of the Welding Journal.

R. D. Thomas Jr. International Lecture

Robert E. Shaw Jr., PE (Fig. 3), presented the 2016 R. D. Thomas Jr. International Lecture titled International Perspectives into Welding Quality for Steel Structures.

Shaw serves as president of the
New Technology

The biggest draw for most show-goers is the products on display. Following are descriptions of some of the products the Welding Journal editors noticed during FABTECH.

Suhner, a company that has been in existence for 102 years and manufactures all its own parts, introduced a series of cordless grinders to fit many different applications. The ASC9 straight grinder, the ABC7 belt grinder, the AKC3 fillet grinder, and the ATC7 tube polisher (Fig. 4) all operate on a rechargeable battery, allowing them to be taken to the job wherever that might be. The battery pack delivers 18 V of power, and it is enclosed in a rubber shield to absorb the hard blows that may be encountered in everyday use. A sensor automatically switches off the machine if it detects overheating or overloading. The ASC9 straight grinder can operate at speeds up to 9000 rpm. The ABC7 belt grinder accepts belts from ½ to ⅜ in. wide. The AKC3 fillet grinder weighs 4.8 lb including a battery pack. The ATC7 tube polisher is designed for tubes up to 1½ in. All of the grinders are packaged in a hard case containing the machine, two battery packs, a charger, and four charging cables that can be plugged into common sockets around the world. Suhner, suhner-abrasive-expert.com

ESAB introduced its 300i Renegade portable power source (Fig. 5) designed for jobs in tight spaces. It weighs 33 lb, and with handles on the rear, front, and top of the machine, it can be easily grabbed for carrying to remote locations. This combination SMAW and GTAW machine is capable of producing 300 A at 40% duty cycle, and it can operate with work leads up to 300 ft. It is fitted with Adjustable Hot Start, a feature that provides optimal energy when initially striking the arc. It accepts covered electrodes up to ¼ in. and is capable of gouging with electrodes up to ⅜ in. There is a function that allows arc adjustment for different electrodes, applications, and individual preference. Up to three preferences can be stored in its memory. ESAB, esab.com

Koike Aronson featured its MD series of positioners. These medium-duty positioners (Fig. 6) can be operated with a remote pendant. The forward tilt is up to 135 deg and capacities range from 1500 to 5000 lb. The clamping diameter ability is 30 to 47...
in. All the models have the versatility to accept plug-in options such as a foot switch that provides control of speeds, rotational axis, and direction. A radio remote control includes a display that lets the operator see system status, diagnostics, and battery life. Koike Aronson, koike.com

The Bosch Rexroth next phase of adaptive resistance welding controller is the **PRC 7000** — Fig. 7. Developed jointly with BMW, the product has multiple processors incorporated in its electronics that allows customization of programs for such factors as different thicknesses and steels. It can store multiple programs. It is also designed to use energy efficiently, with a 30% reduction in energy usage claimed while welding, and 80% less in the idle mode. It has a net current rate of 110 A with a maximum primary current of 550 A. Bosch Rexroth, boschrexroth.com

The **B-500** transportable pipe beveler (Fig. 8) enables an operator to set up and bevel 10-in.-diameter Schedule 40 pipe in about 45 s without flames, replacing traditional beveling methods such as grinding. It can bevel steel and stainless steel pipe 4 in. and greater in diameter and ¾ to ½-in. wall thickness, and steel plate ¾ to ½ in. thick. Other positive characteristics are that it produces a machined finish because of the six cutting inserts that work simultaneously, and the pipe is cool to the touch immediately following beveling. Ridgid, ridgid.com

Fein had on hand its **JHM USA 101**, a precise magnetic-base drill for construction sites — Fig. 10. Features include a uniform-torque motor for heavy-duty core drilling, high concentricity and consistent core ejection (due to special bore spindle guides), solid construction in die-cast aluminum for work under difficult conditions, integrated cooling tank, internal motor cable routing, and hand feed wheel that can be mounted on either side. Fein, fein.com

Riding on the fame of the MPS 630, a robotic tool changer well known in the automotive industry, Stäubli now presents the **MPS 130**, a compact and light model that handles loads up to 220 lb, has a maximum bending torque of 663 ft-lb, and a maximum torsion of 590 ft-lb — Fig. 9. It can be used on applications such as handling, riveting, spot welding, painting, and plastics. Stäubli, staubli.us

Fronius presented the new generation of the TransPocket SMA welding machine. The **TransPocket 180** (Fig. 11) has been designed as a replacement for the TransPocket 1500. The 180 is the first single-phase, 180-A SMAW machine available. The welding torches are designed primarily for SMAW and can use electrodes with diameters up to ¾ in. This latest generation of power source is also suitable for GTAW at up to 220 A, with a new GTA multiconnector and a welding torch with an up/down function to ad-
The TruPunch 1000 enables sheet metal fabricators to upgrade their machine to the **TruMatic 1000** fiber punch laser machine introduced to the North American market at FABTECH 2016 — Fig. 13. A highlight is the modular design that enables the punch machine to transition into the new fiber combination machine. The process is done by adding a 3-kW TruDisk solid-state laser and retrofitting the punch with a laser cutting system, laser evacuation unit, and beam guard system. Both also offer a redesigned drive technology; the patented Delta Drive moves the electric punching head and laser along the y-axis while the sheet moves in the opposite direction. Markets include combination processing and job shops. TRUMPF, us.trumpf.com

**Miller Electric Mfg. Co.** introduced its new **AugmentedArc™** augmented reality welding system for welding education — Fig. 15. The system, which can be used by beginner and intermediate-level welding students, simulates multiprocess gas metal arc welding, gas tungsten arc welding, flux-cored arc welding, and shielded metal arc welding, blending real-world and computer-generated images into an augmented reality environment. Students can practice on all common joint types, including pipe, in all positions. Miller Electric Mfg. Co., millerwelds.com

Weiler touts that its new **Tiger® UltraCut 1 Millimeter** (Fig. 16) thin cutting wheels deliver unmatched cutting performance and toughness with
a true 1 mm thickness. Operators will experience fast, smooth cutting and exceptional control for clean, precise cuts on thin sheet metal, profiles, and small-diameter rods. Other features include reduced burrs and greater stability. Weiler, weilercorp.com

The new high-capacity TR1000 tilt/rotate part positioner from Yaskawa Motoman features 1000-kg payload capacity, tilt speed up to 12 rpm, and rotation speed up to 22.4 rpm — Fig. 17. It also features multiple mounting possibilities and can be paired with a tailstock for longer part handling. Yaskawa Motoman, motoman.com

This was the second FABTECH for Cincinnati, Inc. It displayed its CL-960 fiber laser system (Fig. 18), part of a line of lasers that come in power levels of 2000, 4000, and 6000 W. A key feature of the system is its air-assist cutting capabilities. Dry, clean, compressed air offers a cost-effective gas-assist alternative to nitrogen or oxygen. Since its linear drive motors are not mechanical devices, they offer longer life. The system can cut up to ¾-in.-thick steel. Features include a human/machine interface, easy-to-use touch screen controls, and dual pallet configuration. Cutting tables of 5 × 10, 6 × 12, and 8 × 20 ft are available. Cincinnati, Inc., e-ci.com

The new line of Piranha plasma tables (Fig. 19) was designed to partner with the company’s ironworkers, providing two versatile machines that can fit into relatively small spaces. The plasma tables come in three sizes: C404 (4 × 4 ft), C408 (4 × 8 ft), and C510 (5 × 10 ft). They feature an all-welded frame, servo motors rather than stepper motors on the X and Y axes, automatic torch height control, breakaway torch height control, and a programming system. They offer fully integrated CNC control with a 10-in. LCD screen and an extensive shape library. MegaFab, megafab.com

Norton rolled out the first of its new color-coded RapidFinish™ convolute wheels — Fig. 20. The maroon general-purpose wheels were on display at the show. Heavy deburring wheels in brown and final finishing wheels in green will be introduced soon. The company introduced the colored wheels to provide a visual aid to operators so they know which wheels to use for a particular application. The wheels feature a new grain/bond configuration that offers increased throughput and a new fiber and high-temperature resin bond system that offers 10–30% longer life and a smear-free finish. Norton, nortonabrasives.com

See You Next Year

Don’t miss out on the 2017 FABTECH exhibition. Join your colleagues at McCormick Place in Chicago, Ill., on November 6–9 for this exciting welding and metalworking event.
Shop Talk with Counting Cars

A '69 Cadillac courtesy of Count’s Kustoms, Las Vegas, Nev., attracted a lot of attention in between the welding and metalforming halls — Fig. A. This custom chopper and hot rod dealer is showcased in the reality series Counting Cars. Cast members Kevin Mack and “Horny” Mike Henry also participated in a question/answer session on Nov. 18 guided by AWS Director, Conventions and Meeting Services, Matthew Rubin — Fig. B.

“The key is the bodywork,” Mack added, so when painters get cars, they are perfect. Vehicles are completed in less than three months.

“Don’t think you are going to make a lot of money,” Mack cautioned, because you cannot bill for all the hours you will spend working. “You do it out of love,” he continued, because car backstories are meaningful.

Both personalities divulged how reality TV has changed their lives. “I’m still in awe,” Henry said, noting when the show started he wore a criminal justice ankle bracelet. Now, he enjoys a “complete flip of life.” Mack has put his son through college and gives back to veterans more. However, show timelines need to be followed, so sometimes they need to stay late or even all night to get work finished.

Audience members asked questions afterward. Queries ranged from car restoration advice to finding a specific classic car (hint: online is your best bet). One participant thanked Mack and Henry for positively impacting the younger generation who watch their show and learn that working hard, earning what you get, and learning different skills will take you places.

Students Bring Torches and Cooking Skills to Welding Thunder Competition

The smell of steak, eggs, and competition filled the air over the convention center’s Silver parking lot during the 2016 Welders Without Borders Welding Thunder Competition on Wednesday, November 16, and Thursday, November 17.

The two-day contest sponsored by the American Welding Society’s District 21 challenged 16 high school and college-level welding teams to fabricate and weld a griddle and grill, and then cook a meal off their finished product — Fig. C.

Samuel Colton, a welding professor at Arizona Western College Ernest Lopez Welding Institute in Yuma, Ariz., and the founder of Welders Without Borders and Welding Thunder, was thrilled to bring the fifth annual competition to the FABTECH show for the first time.

“Everyone here is really excited about the opportunity the students have had to travel to Las Vegas to compete and to be able to interact with each other and all of the FABTECH events. They will remember this experience for the rest of their lives,” said Colton.

The first day of competition featured fabrication of the grills. On the second day, students were allowed to
Industry experts judged the welding and fabrication portion of the competition while culinary art students and instructors from Arizona Western College rated the food.

For the second year in a row, Arizona Western College (Fig. D) took home the first-place prize in the fabrication category for postsecondary schools while first-time competitors Paso Robles High School (Fig. E) from Paso Robles, Calif., took home the top spot in the high school category. Winners of all six categories included:

**Welding Fabrication — Post Secondary**
- 1st place — Arizona Western College
- 2nd place — Cuesta College
- 3rd place — Mohave Community College

**Welding Fabrication — Secondary**
- 1st place — Paso Robles High School
- 2nd place — Flagstaff High School
- 3rd place — Northern Arizona Vocational Institute of Technology

**STEM Award**
- Post Secondary — Arizona Western College
- Secondary (tie) — Skyline/Red Mountain High School and Cortez High School

**Best Food Award**
- 1st place — Central Arizona College
- 2nd place — Paso Robles High School
- 3rd place — Cuesta College
FABTECH 2016 presented three panel sessions featuring first-rate professionals who addressed current and important topics to manufacturing.

**State of the Industry: Post-Election Analysis**

This session reflected on the 2016 election results and the impact on manufacturing. Energy might see a lot of benefits: there is a lot of talk about the Keystone Pipeline and there should be more activity with nuclear energy and offshore oil exploration.

The banking sector will likely see some gains because of the possible dismantling of the Dodd-Frank law that was passed in 2010. Revisions to this law could allow mid-sized banks — which are key to the manufacturing industry — to expand more freely, and as a consequence, have a real impact on the growth of the manufacturing industry.

Regarding infrastructure, Chris Kuehl, managing director of Armada Corporate Intelligence, observed: “There has been a lot of talk about infrastructure, which is critical, but we still haven’t seen a concrete plan. The challenge has always been how to pay for it.”

Growth in manufacturing is essential to improving the economy and increasing jobs. Ned Monroe, senior VP of External Relations at the National Association of Manufacturers (NAM), elaborated: “For the last ten months, NAM has been working with the transition teams of all the eligible parties to make sure that they knew that manufacturing is the cornerstone of getting the economy moving again. We are the job creators.”

There might be changes in regulations, such as the Clean Power Plan, coal regulations, and the controversial pipeline projects. Omar S. Nashashibi, partner at The Franklin Partnership, commented: “The President-elect has said he would repeal 70% of the regulation from day one, but to roll back many of the regulations, there must be new regulations in place, and that takes time. There is going to be a focus on executive orders.”

Tax reform — a complicated issue — is high on the agenda, but it’s still too early to know what changes will take place.

With respect to trade, a good portion of the industry depends on our relationship with Canada and Mexico, therefore any changes in NAFTA could have a deep impact. Likewise, what happens with the Trans-Pacific Partnership (TPP) is critical. Kuehl commented: “Just because the TPP dies doesn’t mean that we will stop finding ways to better trade with Asia.” Metal and raw materials are a concern: if the administration decides to increase tariffs on essential metal imports, it could result in significant challenges for many manufacturers.

A recurring message throughout the session was workforce training and promoting education. There still are 345,000 job openings in manufacturing that can’t be filled because of the lack of skilled workers. The panelists proposed actions such as promoting Manufacturing Day, having a portable certification system, expanding hybrid schools that teach manufacturing skills in the last couple of years of high school, and training veterans.

All winning teams received a customized award plaque, Lincoln Electric Power Pack with gloves, gear bag, and a helmet, and a MIG Buddy welding gun holster. The high school welding fabrication team winners also received a Hypertherm Powermax 45 XP® plasma cutting machine.

Additionally, Arizona Western College has offered a $1500 scholarship for the most valuable welder on the winning Paso Robles High School fabrication team. The recipient of the award will be selected by the high school’s faculty.

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**Featured Expert Panel Sessions**

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**People’s Choice Award**
Mohave Community College

**Teamwork Mention**
Post Secondary — Western Maricopa Educational Center
Secondary — Yuma High School
Development Trends in Additive Manufacturing and 3D Printing

Using additive manufacturing technologies can increase manufacturing efficiencies, improve performance, and decrease cost of complex products.

One of the main questions about these technologies still remains: what is the right application? In general terms, if the part requires high volume and is reproducible using more mainstream processes, additive manufacturing and 3D printing probably are not the best solution. But in manufacturing low-volume, complex parts with internal channels and structures that require special materials and extremely close tolerances, additive manufacturing and 3D printing might be the best answer. Now that the technologies and materials have advanced, they are moving toward production and complementing traditional manufacturing.

Jennifer Cipolla, leader of the General Electric Center for Additive Technology Advancement, commented: "GE is using these technologies in the production of jet engine parts such as fuel nozzles, fuel swirlers, and other parts in gas turbines. Additive is also being applied in postproduction processes like cladding, cold spray, and some other hybrid technologies that are coming out."

Cost is a key factor: the typical machine for metals can cost around $1 million. The material or “powder” is much more expensive than the materials used with other processes, and it’s very reactive, so it can only be handled in a facility with the adequate safety protocols. The equipment necessary for all the postproduction processes can add significant costs; companies must plan for the time it will cost to get up and running and the extra cost of designing the parts, which is much more complex than in regular manufacturing.

Additive manufacturing has traditionally been associated with metals, but there have been great advances using plastics, for example, in the medical industry for surgical parts and dental implants. Some of the developments that we will see are the ability to work with mixed materials. This is already being used with plastics — making a piece that combines hard plastic and rubber — but we will start seeing it in metals, and that will open up many possibilities. Likewise, nano-3D printing promises incredible advances in the medical world. In the future, machines will get faster and materials will get cheaper, which will result in lower costs and more accessibility to these technologies.

Advanced Manufacturing: Creating Competitive Advantages for Fabricators

Advanced manufacturing helps the industrial processes through a wide variety of technologies, such as rapid prototyping, cloud software, control systems, real-time data gathering and analysis, advanced robotics, automation, and many other computer-related tools.

A good way for manufacturers to start implementing these technologies is to take a look at product life cycles and determine which stage can first benefit from incorporating a more advanced process or tool. For example, sophisticated, cloud-based software for design, simulation, and analysis is now typically available on subscription plans that can be affordable and can provide a quick jump-start toward more efficient and modern processes.

Lonnie Love, group leader at Oak Ridge National Laboratory, commented: "Start with cloud-based software, because that’s the glue that will help you connect and innovate. Also, find a mentor. Look for companies that have been successful at incorporating these newer technologies. I would recommend hiring young recruits and have them work side by side with older engineers. Their ingenuity brings a fresh and creative perspective on how to solve problems and work with these newer technologies."

By starting gradually, manufacturers can easily implement these technologies at an affordable and sensible pace while moving toward improving their products and services, which will help them stay competitive.

ANDREW CULLISON is publisher, KRISTIN CAMPBELL is features editor, CINDY WEIHL is senior editor, and MARY RUTH JOHNSEN (mjohnsen@aws.org) is editor of the Welding Journal. CARLOS GUZMAN is editor, Welding Journal en Español.
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CERTIFICATION SCHEDULE

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</tr>
<tr>
<td>Biloxi, MS</td>
<td>May 7–12</td>
<td>May 13</td>
</tr>
<tr>
<td>Des Moines, IA</td>
<td>May 7–12</td>
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<tr>
<td>Houston, TX</td>
<td>May 14–19</td>
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</tr>
<tr>
<td>Cleveland, OH</td>
<td>May 14–19</td>
<td>May 20</td>
</tr>
<tr>
<td>Los Angeles, CA</td>
<td>May 14–19</td>
<td>May 20</td>
</tr>
</tbody>
</table>

IMPORTANT: This schedule is subject to change. Please verify your event dates with the Certification Dept. to confirm your course status before making travel plans. Applications are to be received at least six weeks prior to the seminar/exam or exam. Applications received after that time will be assessed a $350 Fast Track fee. Please verify application deadline dates by visiting our website aws.org/certification/docs/schedules.html. For information on AWS seminars and certification programs, or to register online, visit aws.org/certification or call (800/305) 443-9353, ext. 273, for Certification; or ext. 455 for Seminars.

9-Year Recertification Seminar for CWI/SCWI

For current CWIs and SCWIs needing to meet education requirements without taking the exam. The exam can be taken at any site listed under Certified Welding Inspector.

<table>
<thead>
<tr>
<th>Location</th>
<th>Seminar Dates</th>
<th>Exam Date</th>
</tr>
</thead>
<tbody>
<tr>
<td>New Orleans, LA</td>
<td>Jan. 8–13</td>
<td>Jan. 14</td>
</tr>
<tr>
<td>Denver, CO</td>
<td>Feb. 26–March 3</td>
<td>March 3</td>
</tr>
<tr>
<td>Dallas, TX</td>
<td>March 5–10</td>
<td>March 5–10</td>
</tr>
<tr>
<td>Miami, FL</td>
<td>March 5–10</td>
<td>March 5–10</td>
</tr>
<tr>
<td>Sacramento, CA</td>
<td>April 2–7</td>
<td>April 2–7</td>
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<tr>
<td>Charlotte, NC</td>
<td>May 7–12</td>
<td>May 7–12</td>
</tr>
<tr>
<td>Pittsburgh, PA</td>
<td>May 14–19</td>
<td>May 14–19</td>
</tr>
<tr>
<td>Kansas City, MO</td>
<td>June 4–9</td>
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</tr>
</tbody>
</table>

Certified Welding Educator (CWE)

Seminars are given at all sites listed under Certified Welding Inspector. Seminar attendees will not attend the Code Clinic portion of the seminar (usually the first two days).

Certified Welding Sales Representative (CWSR)

CWSR exams are given at Prometric testing centers. More information at aws.org/certification/detail/certified-welding-sales-representative.

Certified Welding Supervisor (CWS)

CWS exams are given at Prometric testing centers. More information at aws.org/certification/detail/certified-welding-supervisor.

Certified Radiographic Interpreter (CRI)

The CRI certification can be a stand-alone credential or can exempt you from your next 9-Year Recertification.

<table>
<thead>
<tr>
<th>Location</th>
<th>Seminar Dates</th>
<th>Exam Date</th>
</tr>
</thead>
<tbody>
<tr>
<td>Seattle, WA</td>
<td>Feb. 27–March 3</td>
<td>March 4</td>
</tr>
<tr>
<td>Houston, TX</td>
<td>March 13–17</td>
<td>March 18</td>
</tr>
<tr>
<td>San Francisco, CA</td>
<td>April 10–14</td>
<td>April 15</td>
</tr>
<tr>
<td>Las Vegas, NV</td>
<td>May 1–5</td>
<td>May 6</td>
</tr>
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<td>Cleveland, OH</td>
<td>June 5–9</td>
<td>June 10</td>
</tr>
<tr>
<td>Dallas, TX</td>
<td>July 17–21</td>
<td>July 22</td>
</tr>
<tr>
<td>Miami, FL</td>
<td>Exam only</td>
<td>Aug. 4</td>
</tr>
<tr>
<td>Kansas City, MO</td>
<td>Aug. 21–25</td>
<td>Aug. 26</td>
</tr>
<tr>
<td>Chicago, IL</td>
<td>Sept. 11–15</td>
<td>Sept. 16</td>
</tr>
<tr>
<td>Pittsburgh, PA</td>
<td>Oct. 9–13</td>
<td>Oct. 14</td>
</tr>
</tbody>
</table>

Certified Robotic Arc Welding (CRAW)

ABB, Inc., Auburn Hills, MI; (248) 391–8421
OTC Daihen, Inc., Tipp City, OH; (937) 667-0800, ext. 218
Lincoln Electric Co., Cleveland, OH; (216) 383-8542
Genesis-Systems Group, Davenport, IA; (563) 445-5688
Wolf Robotics, Fort Collins, CO; (970) 225-7736
On request at MATC, Milwaukee, WI; (414) 456-5454

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Types of Resistance Spot Welds

There are four general categories of resistance spot welds: direct, parallel, series, and indirect. The differences are mainly due to the current path through the weldment and the arrangement and shape of the electrodes. Following are descriptions of each of the four types. Keep in mind that while the descriptions and illustrations of the various spot welding processes are generally given with the work surface in a horizontal plane, in actual practice, the work can be in any plane if the axis of the tip face and the direction of the welding force are both normal to the work surface.

**Direct Welds.** In a direct weld, the current path is directly through the work between opposed electrodes. Figure 1 shows two types of direct welds. Figure 1A is a basic spot weld. It is the easiest weld to produce since the electrodes can be of optimum diameter and face contour. Most welding schedules are based on this type of weld.

**Parallel Welds.** In this variation, two or more direct welds, usually closely spaced, are made simultaneously with the welding current for all welds supplied by one transformer. This requires all of the elements involved in the spot welding process to be nearly equal for each weld. Of these elements, the welding current is the most difficult to equalize.

Parallel welding has two advantages when compared with the same welds in sequence. One advantage is that multiple welds can be made in the same time interval required for one weld. The other advantage is that the shunting current through adjacent welds is less, thus usually allowing the minimum weld space to be reduced.

As mentioned previously, the welding currents for each weld must be nearly equal. Better balance can be achieved by using a transformer with multiple secondary windings with each winding feeding one circuit, as shown in Fig. 2.

**Series Welds.** Figure 3A, B shows two variations of series welding. The secondary circuit is connected to two contoured electrodes that contact the workpiece from the same side and a current-conducting backup mandrel is used. Note that a weld is made by each electrode.

Figure 3A shows a basic series welding circuit, which has three parallel current paths between the electrodes. Figure 3B shows a series weld in which the upper workpiece is not continuous between the electrodes, thus eliminating the shunting current.

The principal advantages of the series welding circuits in Fig. 3A, B are the ability to make welds with the electrodes and secondary circuit on only one side of the weldment and the absence of objectional marking on the opposite surface. However, when compared to direct welding, the electrode marking is more severe and the heat balance is not as good due to the flat surface of the mandrel.

Figure 3C illustrates push-pull welding, another variation of series welding.

**Indirect Welds.** Figure 4A–D shows a number of indirect welding arrangements. Figure 4A is identical to the circuit of Fig. 3A for series welding except only one of the electrodes is contoured. Thus, only one weld is made rather than two as is done in the series circuit. Figure 4C shows welding of a cinch flange. Figure 4D is an indirect push-pull arrangement. This is the same as Fig. 3C except that only one set of opposing electrodes is contoured, thereby making only one weld.
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Fiber Laser Welding of WC-Co and Carbon Steel Dissimilar Materials

This investigation focused on the factors that influence the strength and ductility of dissimilar joints

BY P. XU, D. ZHOU, AND L. LI

ABSTRACT
Welding parameters were investigated for fiber laser welding of cemented carbide WC-Co and steel dissimilar materials. The microstructure, composition, phase, and bend strength of the joints were analyzed using optical metallography, scanning electron microscopy, x-ray diffraction, transmission electron microscopy, and bend testing. The optimized welding parameters included laser power of 2 kW, scanning speed at 0.96 m/min, and heat input of 125 J/mm. The flexural bend strength and yield strength of the joints attained 970 MPa and 876 MPa, much higher than that of conventional brazed joints. The brittle fracture during bending occurred along the fusion boundary and HAZ on the cemented carbide side, where dissolution of WC and penetration of Fe from the fusion zone are believed to have caused embrittlement at the WC-matrix interfaces.

KEYWORDS
• Fiber Laser • Dissimilar Joints • Cemented Carbide • Brittle Fracture

Introduction
Cemented carbide is a composite material of hard carbides and a soft binder metal. The hard carbides include those of Group V and Group VI elements, such as WC, TiC, Mo2C, TaC4, Cr3C2, VC, and NbC. The soft binder metal is usually cobalt, nickel, iron, or their mixture (Ref. 1). Cemented carbide has been widely used in aerospace, electronics, marine, petrochemical, mining, and automotive industries for decades in engineering applications, such as pipe-valve components, cutting tools, catalytic converters, rock drill tips, and various wear-resistant parts (Refs. 2, 3). Several methods have been used for joining cemented carbide to steel substrates for drill bits and cutting tools. These methods include brazing (Refs. 4–6), sinterbonding (Ref. 7), or diffusion bonding (Refs. 8–10), chemical vapor deposition (Ref. 11), tungsten arc welding (Ref. 12), friction welding (Ref. 13), and, more recently, laser welding (Refs. 14–17). Okita et al. (Ref. 13) adapted friction welding to join the cemented carbide and steel. A tool steel was friction welded to cemented carbide with an intermediate layer that was dispersion strengthened by tungsten carbide particles in a nickel matrix. The joint tensile strength was found to be greater than 730 MPa when the forge pressure was lower than 250 MPa, but the strength markedly decreased when the forge pressure was greater than 300 MPa.

Tian et al. (Ref. 14) studied dip soldering and welding of carboloy and steel dissimilar materials. Combining laser fusion welding with dip soldering, they found fewer fissures in carboloy, and laser fusion welding could suppress polycrystalline formation in the binding Co, and thus improve the welded joint toughness. Research by Barbatti et al. (Ref. 17) indicated that laser beam welding allowed the successful autogenous joining of a steel to cemented carbide. By welding with a preheating and postweld heat treatment, the temperature gradients were controlled, and lower residual stress level, crack-free, and nonporous, joints were obtained. The mechanical properties of the joints were found to be comparable with those of the conventional brazed steel-cemented carbide joints. Costa et al. (Refs. 15, 16) found the major defects during laser welding strength can achieve 370 MPa.

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The challenge in joining cemented carbide and steel dissimilar materials is the low strength and poor ductility of the metallic bond. The strength of brazed joints usually ranges from 150 to 250 MPa (Ref. 5). With special brazing filler metals and pretreatment of cemented carbide, the brazed joint strength can achieve 370 MPa. Okita et al. (Ref. 13) adapted friction welding to join the cemented carbide and steel. A tool steel was friction welded to cemented carbide with an intermediate layer that was dispersion strengthened by tungsten carbide particles in a nickel matrix. The joint tensile strength was found to be greater than 730 MPa when the forge pressure was lower than 250 MPa, but the strength markedly decreased when the forge pressure was greater than 300 MPa.

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of WC-12%C cemented carbide to 0.25%C steel to be misalignment, porosity, cracking, and excessive melt-through (for a specimen thickness of 2.5 mm). The horizontal position of the laser beam interaction area was identified to be a major factor for joint quality. The quality was optimized when the laser beam was positioned on the steel side with a distance of 0.2 mm from the bond centerline. If the laser beam interaction area was placed closer to the hard metal, metal cracking was easily observed. However, if the laser beam was positioned farther from the joint (greater than 0.2 mm), the parts did not fully join due to incomplete fusion. The microstructure of the fusion zone was found to be a cellular dendritic structure with an eutectic mixture of austenite and complex carbides occupying the interdendritic spaces of primary austenite dendrites. However, the strength and ductility of laser joints as influenced by dissimilar material welding mechanism have not been studied.

With recent advancements of readily accessible and efficient lasers, especially fiber laser technology, the time has come to make laser welding competitive relative to brazing for achieving strong and possibly ductile joints. This paper provides the results of an investigation into the process parameters for fiber laser welding of WC-20Co to AISI 1045, both popular materials for engineering applications. The focus was on the mechanism for joining and the factors that influence the strength and ductility of the dissimilar joints.

**Experimental Procedure**

**Materials and Welding Procedure**

The WC-20Co cemented carbide was used as one of the base materials. The alloy has the following chemical composition: 4.9C, 20Co, and 75.1W (wt-%). A carbon steel, AISI 1045, was used as the other base material. The carbon steel has the following chemical composition: 0.45C, 0.28Si, 0.62Mn, 0.004S, 0.004P, 0.25Cr, 0.25Cu, and balance Fe (wt-%). Disc-shaped base materials with a 50 mm diameter and three thicknesses of 2, 3, and 4 mm were brushed to a 2-μm surface roughness finish. These base metal discs were cut into halves along the diameter and clamped to form a cemented carbide to carbon steel butt joint with no root opening — Fig. 1. A 5-kW maximum output YLS-5000 fiber laser (IPG, USA), with a KR60-HA robot system (Kuka, Germany), and a BIMO QBH laser processing head (HIGHYAG, Germany), was used to weld the butt joint. During the autogenous welding, a copper backing strip was used to support weld root formation. The weld coupons were rigidly clamped to obtain low angular distortions of the joints. The laser beam focal point was varied from 0 mm (on the surface of the plate) with a spot radius of 0.1 to –10 mm defocusing amount. A front and back shielding, provided by gas trailers, was supplied with a high-purity argon gas at a flow rate of 15 to 25 L/min to prevent the molten pool and heat-affected zone (HAZ) from oxidation. The process parameters for laser welding are shown in Table 1. Following welding, the welded joints were evaluated for bead formation, incomplete fusion, microcracking in the fusion zone, or possible liquation cracking in the WC-Co side of the HAZ.

**Bend Test and Microstructure Analysis**

Three-point bend strength was measured using a Zwick BTC-T1-FR020TN.A50 universal testing frame (Zwick, Germany) that is stepper motor driven and with a 20-kN load cell. The welded specimens were cut into bend test coupons with dimensions of 48 × 4 × 2 mm (specimens A2-2, A2-8, and A2-9) or 48 × 4 × 3 mm (specimens B3-2). The as-welded surfaces were ground along the longitudinal axis of the test coupons. The length of the three-point bend test span (L) was 36 mm. All coupons were tested in face-bend configuration at ambient...
Table 1 — Laser Welding Parameters and Formation of Welds

<table>
<thead>
<tr>
<th>Specimen</th>
<th>Specimen thickness (mm)</th>
<th>Defocus (mm)</th>
<th>Preheat Laser Power (kW)</th>
<th>Welding Laser Power (kW)</th>
<th>Scan Speed (m/min)</th>
<th>Energy Input (J/mm)</th>
<th>Complete Weld Penetration</th>
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</thead>
<tbody>
<tr>
<td>A0-1</td>
<td>2</td>
<td>0</td>
<td>0</td>
<td>2.7</td>
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<td>0</td>
<td>2.7</td>
<td>1.8</td>
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<td>-8</td>
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<td>2.7</td>
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<td>2.0</td>
<td>1.82</td>
<td>118</td>
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B0-1     | 3                       | -8           | 0                        | 3.0                      | 1.20              | 150                | Yes                    |
B0-2     | 3                       | -8           | 0                        | 2.7                      | 1.08              | 150                | No                     |
B1-1     | 3                       | -8           | 0                        | 3.0                      | 1.08              | 166                | Yes                    |
B1-2     | 3                       | -8           | 0                        | 2.7                      | 0.98              | 166                | No                     |
B2-1     | 3                       | -8           | 0                        | 2.7                      | 0.84              | 192                | Yes                    |
B2-2     | 3                       | -8           | 0                        | 2.5                      | 0.78              | 192                | Yes                    |
B3-1     | 3                       | -8           | 0                        | 2.7                      | 0.78              | 208                | Yes                    |
B3-2     | 3                       | -8           | 0                        | 2.5                      | 0.72              | 208                | Yes                    |
C0-1     | 4                       | -8           | 0.5                      | 3.0                      | 1.56              | 115                | No                     |
C0-2     | 4                       | -8           | 0.5                      | 3.5                      | 1.56              | 135                | Yes                    |
C0-3     | 4                       | -8           | 0.5                      | 3.7                      | 1.56              | 142                | Yes                    |

The microstructure of the laser welds was characterized by x-ray diffraction (XRD), optical metallography, scanning electron microscopy (SEM), and transmission electron microscopy (TEM). The XRD measurements were carried out on weld samples using an X’Pert PRO X-ray diffractometer with a Cu Kα radiation (λ = 0.15406 nm) and a BLK2 cooling cycle system. The scanning step size was 0.026 deg, and the scanning range was 15 to 120 deg continuous. The scanning speed was 0.438 deg/s. For the XRD, the current and voltage were at 40 mA and 40 kV, respectively.

The optical microscopy and SEM specimens were prepared by mounting, grinding and polishing, and etching with the Murakami’s reagent (10 g potassium ferricyanide K3Fe(CN)6, 10 g sodium hydroxide NaOH, and 100 mL water, freshly prepared) (Ref. 27). High-resolution microstructure of the as-welded fusion zone was characterized using a JEM 2100 TEM (JEOL, Japan). Samples for TEM were prepared using ion milling. The TEM parameters were 200-kV acceleration voltage, 96–120 kV, and 109.8-pA/cm2 current density, 2-s exposure time, and magnifications between 20k and 200k times.

Results

Weld Formation

The weld penetration and bead formation are affected by welding parameters (Table 1). The laser spot position was found to influence the fusion of the dissimilar materials. As the melting point of WC (approx. 2700°C) is much higher than that of carbon steel (approx. 1350°C), the laser spot needs to be located at 1 mm from the butt joint line toward the cemented carbide...
side. If the laser spot focuses at the steel side (specimen A0-1) or on the butt joint line (specimens A0-2, A0-3, and A0-4), incomplete fusion was shown to have happened on the WC side. Among the welding parameters, the defocusing amount was found to be a sensitive factor to influence the weld formation of WC-20Co cemented carbide to AISI 1045 carbon steel. The results indicated that specimen A0-3 (defocusing amount –8 mm below surface), in contrast with specimen A0-2 (defocusing amount was zero, or on surface), obtained complete joint penetration without weld spatter. Therefore, all subsequent welding trials were conducted with a –8 mm defocused laser spot on the WC-Co side.

Local preheating by a "dry-run" of the laser scan (i.e., all parameters kept the same, except the laser power being reduced to 1/10th of the level for welding) was included in the welding procedure (Table 1). Figure 2 illustrates the influence of preheating on the weld formation during fiber laser welding. The specimens with preheating (specimens A1-1, A1-2, and A1-4) had consistently better penetration than those without preheating (specimens A2-1, A2-4, and A2-8). However, microcracks were observed on the surface of the cemented carbide HAZ if the preheat is combined with an increase of welding heat input (e.g., specimen A1-1).

The relative effect of laser power and scan rate for a constant heat input on weld formation was investigated (Table 1). With a heat input of 150 J/mm, if the laser power is 2.7 kW and scan rate is 1.08 m/min (specimen B0-2), a good weld formation on the front side but a poor penetration of weld bead on the back side were obtained for the 3-mm-thick specimens. While the heat input was kept constant at 150 J/mm, if the laser power was increased to 3.0 kW and scan rate increased to 1.20 m/min (specimen B0-1), complete penetration with good weld formation was obtained. A similar result was obtained for a constant heat input of 166 J/mm. With a laser power of 2.7 kW and scan rate of 0.98 m/min (specimen B1-2), insufficient penetration was observed at the end of the weld; while complete joint penetration was obtained with an increased laser power to 3.0 kW and a increased scan rate of 1.08 m/min (specimen B1-1). These results can be observed in Fig. 3, which shows the weld formation and penetration of specimens B0-1, B0-2, B1-1, and B1-2. Within the range of parameter variations in this study, it seems that for a constant heat input, increasing the laser power has a greater effect than decreasing the laser scan rate, on enhancing the penetration.

Figure 4 shows the changes in weld formation for specimens B2-1, B2-2, B3-1, and B3-2, when the heat input was increased from 192 J/mm to 208 J/mm. The weld widths on the front side are 2.39, 2.18, 2.5, and 2.29 mm, and the widths on the back side are 1.26, 1.38, 1.2, and 1.31 mm, respectively. No cracks were observed in complete-joint-penetration laser welds except on specimen B3-1. Therefore, for 3-mm WC-20Co and AISI 1045 steel laser welding, it is recommended to use a heat input in the range of 192 to 208 J/mm, and a laser power in the range of 2.5 to 2.7 kW.

With the increase of plate thickness to 4 mm, it becomes difficult to join the WC-20Co cemented carbide to AISI 1045 steel using the current welding setup. Poor penetration with
unacceptable weld formation was obtained if the heat input was below 115 J/mm. When the heat input was increased to above 134 J/mm, not only longitudinal microcracks, but also the transverse microcracks were observed on the surface and cross sections of the laser fusion zones.

Microstructure of Joints

A typical joint has the nail-head shaped autogenous weld fusion zone with a smooth top and root formation — Fig. 5. There is a greater dilution from the steel side than from the WC-Co side. A micrograph of the fusion boundary region on the WC-Co side is shown in Fig. 6. The fusion zone on the left side of the figure appears to have solidified in primary dendritic and eutectic microstructure; the cemented carbide base material on the right side of the figure appears to have retained the cubic and triangular WC particles in the Co matrix. The fusion zone close to the AISI 1045 side fusion boundary also appears to have a primary dendritic and eutectic microstructure, but seems to show an increased size for primary dendrites — Fig. 7. The center of the weld fusion zone shares a similar primary dendritic and eutectic microstructure — Fig. 8. A chemical analysis of the key points in the microstructure revealed the composition of a 0.7 wt-% carbon steel alloyed with 15 wt-% W and 5 wt-% Co (Table 2). It is evident from the dark needles that the steel dendrites may have further transformed to martensite-austenite constituents on-cooling — Fig. 8. The eutectic regions have a typical composition of 50 wt-% Fe, 45 wt-% W, and a relatively higher C concentration; therefore, it is reasonable to suggest one eutectic phase to be $\text{W(Fe)}_2\text{C}$ carbide (Table 2).

Specimens from the weld fusion zone were analyzed by XRD for crystal structures — Fig. 9. The phases identified in the fusion zone included $\alpha$-ferrite, martensite, Fe-containing carbides ($\text{Fe}_3\text{W}_3\text{C}$ and $\text{Fe}_5\text{W}_6\text{C}$), and a small amount of $\text{MC}$ ($M$ being W and Fe) carbide. Due to the overlapping peaks of ferrite and martensite, they remain to be differentiated. However, the high-carbon content in the dendritic regions seems to favor the identification of the peaks due to martensite. If ferrite was present at room temperature for this fusion zone, the extra carbon would have precipitated as carbides, because the solubility of carbon in ferrite was low. There is no evidence of carbides in the primary dendritic regions under the resolution of the BSE micrograph — Fig. 8.

The phases in the fusion boundary and HAZ of the WC-Co base material were identified from XRD of longitudinal specimens extracted parallel to the welds — Fig. 10. The phase composition for the HAZ is similar to that of the weld fusion zone, including WC, some eutectic carbides ($\text{Fe}_3\text{W}_3\text{C}$ and $\text{Fe}_5\text{W}_6\text{C}$), and a weak indication for $\alpha$-Fe (ferrite) and martensite. The dominant phase in the HAZ was WC carbide, which was not significant in the fusion zone. It was notable that the Co binding phase (of the HCP crystal structure) in the base material WC-Co was not detected in the HAZ. Significant alterations in the Co binding phase in the WC-Co HAZ must have happened during laser welding.

Bend Strength and Fractography

Due to the statistical nature of mechanical properties of cemented carbides, four repeat specimens were bend-tested for each welded sample. Figure 11 summarizes the flexural stress-flexural strain curves from the three-point bend testing of typical
joints. For 2-mm-thick sample A2-2, which was welded with a high heat input, all four specimens exhibited a linear stress-strain curve, the bend strength falling between 311 and 508 MPa, without showing any ductility — Fig. 11A. With the same heat input (achieved using an increased laser power and increased scan speed), 2-mm-thick sample A2-8 exhibited not only a much higher average bend strength (about 826 MPa), but also significant ductility — Fig. 11B. Compared with sample A2-8, a decreased heat input for 2-mm-thick sample A2-9 resulted in three of the specimens to exhibit in low strength (375 MPa average) and one specimen exhibiting high strength and ductility — Fig. 11C.

Figure 11D shows the flexural stress-strain curves of 3-mm-thick specimen B3-2. The maximum and minimum bend strengths were 844.31 MPa and 318.72 MPa with the range of plastic deformation being 0.12 to 0.47 mm.

Table 3 lists the flexural strength of the tested specimens. As can be seen, the maximum bend strength using optimized welding parameters is 970 MPa, and the minimum bend strength is 312 MPa for 2-mm-thick specimens. Even the minimum bend strength compares favorably with reported typical strength for brazed joints.

Except for a few cases in which the specimen contains hot-cracking defects in the fusion zone and the fracture happened in the fusion zone, all bend tested specimens fractured along the fusion boundary on the WC-Co side. An example is shown in Fig. 12, in which the face-bend brittle fracture is shown to have started in the weld fusion zone near the weld toe, propagated across the fusion boundary, and grown parallel to the fusion boundary, but always within the HAZ. The fractography of a typical fractured surface is shown in Fig. 13. In the weld fusion zone portion of the fracture, the fracture mode is intergranular — the fracture separates the interdendritic boundaries and reveals the tips of dendritic arms and the eutectic constituents — Fig. 13A. In the HAZ portion of the fracture, the fracture mode is transgranular — Fig. 13B. At a higher magnification, the fracture in the HAZ showed the undissolved WC particles separated in cleavage mode, and the binding phase separated in microvoid coalescence (dimple) mode, which showed some local ductility — Fig. 14. Near the right center of the micrograph, the arrow points to a cluster of as-solidified dendritic tips, which suggest the existence of mi-

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**Table 3 — Flexural Bend Strength of the As-Welded Specimens**

<table>
<thead>
<tr>
<th>Specimen ID</th>
<th>Elastic Modulus (GPa)</th>
<th>Yield Strength (MPa)</th>
<th>Tensile Strength (MPa)</th>
</tr>
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<tr>
<td>A2-2-01</td>
<td>256.94</td>
<td>–</td>
<td>473.05</td>
</tr>
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<tr>
<td>A2-2-03</td>
<td>224.24</td>
<td>–</td>
<td>440.46</td>
</tr>
<tr>
<td>A2-2-04</td>
<td>246.98</td>
<td>–</td>
<td>311.60</td>
</tr>
<tr>
<td>A2-8-01</td>
<td>249.28</td>
<td>731.71</td>
<td>867.59</td>
</tr>
<tr>
<td>A2-8-02</td>
<td>285.97</td>
<td>–</td>
<td>701.21</td>
</tr>
<tr>
<td>A2-8-03</td>
<td>267.20</td>
<td>875.96</td>
<td>970.06</td>
</tr>
<tr>
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<td>251.47</td>
<td>–</td>
<td>686.20</td>
</tr>
<tr>
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<td>A2-9-02</td>
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<td>842.88</td>
<td>880.01</td>
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<tr>
<td>A2-9-03</td>
<td>271.14</td>
<td>–</td>
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<tr>
<td>A2-9-04</td>
<td>269.26</td>
<td>–</td>
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<tr>
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<td>B3-2-04</td>
<td>191.14</td>
<td>834.70</td>
<td>844.31</td>
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</tbody>
</table>
Crofissures due to hot tearing near the fusion boundary.

Discussion

The dissimilar materials were fusion welded autogenously, i.e., without adding a filler metal. An analysis of cross-sectional images of joints enabled an estimation of the fusion ratio for the two base materials. Of the welds in this study, on average 20% of the typical fusion zone was melted in from the WC-Co cemented carbide side, and 80% was melted in from the AISI 1045 side. The weld fusion zone chemical composition was thus estimated as 1.2 wt-% C, 0.2 wt-% Si, 0.6 wt-% Mn, 82.2 wt-% Fe, 12.5 wt-% W, wt-% C, and 3.3 wt-% Co. An Fe-W pseudo-binary phase diagram was calcu-
lated for this composition using the TCFE6 database of Thermocalc — Fig. 15. The formation of the fusion zone microstructure upon solidification can be understood using the diagram. For a 12.5%W fusion zone composition, the first solid phase to appear on solidification is the primary dendrites of austenite (\(g\)), then there is a multi-component reaction near 1500 K to form M\(_6\)C + \(\mu\) phase + eutectic. The \(\mu\) phase is a Fe-W intermetallic. Upon further cooling, the austenite (\(g\)) would transform to a mixture of ferrite (\(a\)) + MC carbide + graphite near the 1000 K eutectoid region. Since rapid cooling during laser welding suppresses the diffusional eutectoid reaction, the austenite will transform diffusionlessly to martensite instead. This proposed sequence of transformations seems to be supported by the microstructural, chemical, and phase identifications presented above.

For a composite material, such as WC-Co, the formation of fusion boundary can be complicated because of the different tendencies of dissolution and melting of the two components in the composite. The current case seems to show the fusion boundary region to be a boundary region of roughly 30 micrometers thick — Fig. 16. The left edge of the boundary region is the border line distributed on which are partially melted WC carbides. The right edge of the boundary region is indicated by the arrows beyond which there is no evidence for binder melting. In a different specimen, the structure of the fusion boundary region shows eutectic features in the binding phase several grains away from the left weld interface — Fig. 17. The arrows point to the extent in which the eutectic features can still be observed in the binding phase. Therefore, for WC-Co composite, the fusion boundary is a region defined on the higher temperature side by melting of the higher melting point component (WC), and on the lower temperature side by melting of the lower melting point component (Co).

The significance of identification of this fusion boundary region instead of a weld interface is that some earlier observations can now be explained. In the XRD results of both the fusion zone and the HAZ, no original Co binding phase was found. Within this fusion boundary region, the absence of Co can be readily explained by the penetrating Fe and dissolving W that altered the hexagonal Co binding phase to a Fe-based alloy (a cubic crystal system). This proposed boundary region also explains why the HAZ on the WC-Co side is the weakest link for fracture in bend tests. Original more ductile Co was replaced with high-carbon, W- and Fe-enriched brittle eutectic constituents.

TEM analysis is conducted on the HAZ microstructure to verify the dissolution of WC into Co. Figure 18 shows the bright field image of the HAZ of sample A2-8. Two dissolving WC particles are surrounded by the Co binding phase. Zooming in on the point indicated by the arrow, under high-resolution TEM, the WC-Co interface structure is shown in Fig. 19. First, the dissolving WC particle actually turned to \(W_2C\) structure at the interface, as observed by other researchers (Ref. 18). The d-spacing of \(W_2C\) (1011) planes is 2.229 angstrom. The d-spacing of (1121) planes in the -Co matrix near the \(W_2C\) interface has two values, 2.092 and 2.109 angstrom, with the higher value closer to \(W_2C\). This indicates that through solid solution, the dissolved W and C changed the lattice parameter of Co near the \(W_2C\) and \(-\)Co interface.
Conclusions

Three-mm-thick WC-Co and steel dissimilar materials were successfully welded using fiber laser welding. The optimized welding parameters include laser power of 2 kW, scanning speed at 0.96 m/min, and the heat input of 125 J/mm. The flexural bend strength and yield strength of the joints attained 970 MPa and 876 MPa, much higher than that of conventional brazed joints. Brittle fracture during bending occurred along the fusion boundary and HAZ on the cemented carbide side, where dissolution of WC and penetration of Fe from the fusion zone are believed to have caused embrittlement at the WC-matrix interfaces.

The joint formation in a WC-Co composite seems to involve the formation of a fusion boundary region that is several WC grains wide (approximately 30 micrometers). The higher temperature border of the fusion boundary region is defined by the melting point of WC. The lower temperature border of the fusion boundary region is defined by the melting point of the Co binding phase.

Acknowledgments

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References

WELDING RESEARCH

Fig. 18 — TEM bright field image of the WC-Co HAZ.

Fig. 19 — HRTEM image of the carbide matrix interface indicated by the arrow in Fig. 18.


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Introduction

Motivation

The present work is a continuation of an initially imposed engineering task — altering an excessive wettabili-
ty of Mo0.4Ni0.6 brazing filler metal (BFM) used for bonding refractory metals, e.g., molybdenum, rhenium, and/or porous tungsten (Ref. 1). New BFM were made by adding nanoscale SiC particles to the microsized Mo0.4Ni0.6 powder. The Mo0.4Ni0.6 wet-
tability change was achieved by adding nanoparticle of the same material (Mo-Ni) or a distinct material, such as SiC (Ref. 2).

In the latter case, a marked difference in the joint microstructure was noticed after solidification of the BFM liquid phase (Ref. 3). However, these microstructural changes were not analyzed at the time of discovery. This led to the hypothesis that refinement of the microstructures of solidified Mo-Ni-SiC using nano-micro powder composite brazes may lead to a better mechanical integrity of the brazements (Refs. 4, 5). Hence, a more detailed study of this phenomenon would be warranted.

This paper offers a detailed account of these observations. More specifically, it addresses a shear stress improvement upon doping due to microstructural modifications, and offers a detailed discussion of the reasons for the microstructural changes induced by adding SiC nanoparticles to the eutectic Mo-Ni powder.

State of the Art

Brazing is used in many molybde-
um applications (Refs. 6–8). Unfortu-
ately, low-melting brazing filler met-
als, such as BAg-series alloys, cannot be used at high temperatures when Mo is predominantly utilized. As a re-
response to Mo service conditions, high-
temperature BFM such as V-35Nb, V-
Mo (Ref. 9), Ni-14Cr-6Fe (wt-%) (Ref. 7), Ti-based alloys (Refs. 6, 8), Fe-15Mo-5Ge-4C-1B (Ref. 10), and Ni-Mo eutectic alloys (Ref. 7) were developed and have been used. The Ni-14Cr-6Fe (wt-%) filler metal yields joints with a rather low, 132-MPa, shear strength (Ref. 8).

Transient liquid phase (TLP) pro-
cessing, in which a thin layer of Ni is used for diffusion bonding, requires a long temperature dwell that could cause undesirable molybdenum embrittlement due to the recrystalliza-
tion process (Ref. 9). The binary Mo-
Ni BFM has been used for a long time in refractory metal bonding (Ref. 7).

Recently, Mo-Ni BFM with eutectic composition was exploited in joining molybdenum to porous tungsten for microwave technology applications (Ref. 11). The filler metal powder was produced using elemental powder milling. Brazing was carried out at 1400°C for 15 min. The resulting joint tensile strength was rather low, about 88 MPa. No detailed description of the joint microstructure was presented.

In our study, we decided to explore the behavior of Ni-Mo eutectic BFM doped by SiC. We aimed at examining the mechanical integrity of the doped
joints. To achieve that objective, we performed two distinct sets of experiments with the same set of filler metal systems (Mo-Ni + x% SiC). The first was a shear stress testing of the brazed Mo-Mo lap joint samples of a distinct geometry. The second was a detailed microstructural study of a semi-infinite brazed joint involving molybdenum as the base metal substrate.

Experimental

Materials and Procedures

Brazing filler metal preparation. All BFM used in this work were manufactured from a binary Mo-Ni micron-scale powder as their major base component. The powder had 34.71 at.% Mo and 65.28 at.% Ni (46.5/53.5 wt-%) composition and <350 mesh dimensions (40–150 μm particle size). The first specimen material designated as 4-2a (Table 1) consisted only of Mo-Ni powder and served as the baseline for the consideration of metallurgical effects on joint properties via additions of SiC powder to the binary Mo-Ni powder. The SiC powder particle dimensions were in the nano range, i.e., <50 nm. Three doped powder mixtures were prepared via additions of 1, 3, and 5 wt-% (2.44, 6.96, and 10.88 at-%) of SiC to Mo-Ni microscale size powder.

Differential thermal analysis and alloy melting characteristics. Differential thermal analysis (DTA) was carried out on a NETZSCH STA 449 C instrument. The alloys’ melting temperatures were determined from cooling segments of DTA of Mo-Ni + SiC alloys. The considered alloys were heated up to 1460°C. It was observed that the crystallization temperature (T$_c$) monotonously decreased with the SiC addition, from 1310°C for binary base Mo-Ni alloys containing no SiC addition to 1285°C for alloys containing 5 wt-% SiC. This was expected from the well-known impact of metalloids on transition metals, i.e., decreasing transition metals’ melting temperatures (observed in our DTA measurements) and decreasing surface tensions of these alloys in the liquid state (Ref. 12).

Joint Integrity Testing Procedures

Shear stress testing. Mechanical evaluation was made using the two-dimensional pure shear mode testing method developed by Rabinkin and Pounds (Ref. 13). In our case, we prepared specimens consisting of two rectangular molybdenum blocks with 10 x 10 x 50 mm linear dimensions, which were machined from solid Mo rods. The blocks were placed in a fixture in which they were tack welded to the supporting plates in the lap-type configuration in which a 50 μm brazing gap between molybdenum blocks was secured — Fig. 1A.

After the brazing operation, the tack-welded plates (i.e., fixers) and an excessive amount of the solidified filler metal in the fillets were machined off the brazed samples to create a well-defined rectangular cross-section geometry of both the specimens and brazements. Figure 1B shows samples after brazing and machining operations together with the appearance of brazed samples after mechanical testing.

To compare the strength of the joint with virgin Mo base metal, the step-like specimens (Fig. 1C), so-called Z-specimens, were machined from the solid Mo block to make their geometry similar to that of the brazed ones. These Z-specimens were placed in the furnace together with specimens to be brazed during brazing operation to get
the same state of the virgin Mo as in the brazed samples.

**Semi-infinite Mo/BFM joint configuration for metallography.**

Mo/BFM joints convenient for metallographic studies were manufactured from caps made of pure molybdenum, and each was loaded by BFM powders with different compositions. These loaded specimens were subjected to the same heat treating cycle as those made for shear testing. Namely, the loaded caps were placed together in a controlled atmosphere furnace (75% H₂ + 25% N₂) and heated until the powders within the caps melted. During heating, the specimens were at temperatures above 1319°C for about 1 min with the temperature peak achieving 1362°C. A fast cooling to room temperature followed.

All four specimen materials were then used to study metallurgical reactions between Mo and Mo-Ni + SiC at the joint interface and the microstructure of a brazing filler metal in the interface vicinity and beyond. They were cut in half, imbedded in the standard 31-mm-diameter resin cylinders, and subjected to the following multistage polishing procedure.

First, specimens were polished using different silica sandpapers followed by diamond suspension polishing. Lastly, Struers OPU paste was applied to finish the preparation of samples.

**Joint mechanical integrity testing.** Two types of mechanical integrity testing procedures followed. These included shear tests as well as nano-indentation tests. Due to space limitation, only the former will be discussed, with the latter addressed briefly and elaborated elsewhere.

The shear testing specimen guide fixture was placed on the Instron 5582 machine table to provide lateral support to specimens in the lap-type configuration, thus avoiding formation of the three-dimensional stress field during vertical loading (Ref. 13). The loading of all tested specimens was applied with 1.25 mm/min rate.

Nano-indentation experiments were performed using atomic force microscopy (AFM) to characterize the relative hardness of phases in the Mo/Mo-Ni brazed joints that were modified by the addition of SiC nanoparticles.

AFM has been used so far not only for the specimen’s surface characterization but also for assessment of the elastic modulus of some polymer fibers and creep evaluation (Ref. 14). Traditionally, AFM utilizes a fine tip probe mounted at the end of a cantilever to measure the surface morphology. In our case, for hardness assessment, we used Veeco (Digital Instruments) Dimension 3100 AFM instrumentation. For this instrument, a special diamond tip was mounted to a stainless steel cantilever. This diamond tip was pressed into the surface of a sample with a known force to create a nano indentation. The cantilever force used in this study was 1.73 × 10⁻⁵ N.

For each indent, a plot of the cantilever deflection versus the Z displacement (called a force plot) was created. An image of the indent was recorded by SEM. Afterward, the depth of the indent was determined by performing a cross-section scanning analysis through the center of the indent with the diamond tip movement used to perform this scanning. It was proven by the same method of measurements on standard gold and molybdenum specimens that the phase hardness on the nano scale is inversely proportional to the depth of the imprint. Therefore, the comparative hardness of our alloy phases can be expressed well via corresponding indent depth.

Because silicon and carbon were found to be nonuniformly distributed among the Mo/Ni phases, the AFM nano-indentation measurements characterize the role of the silicon and carbon as active factors changing the joint microstructure morphology, and as potentially hardening components in the joints’ phases.

Metallographic studies were made using optical microscopy (OM) and SEM performed on a JEOL JXA 840 electron microprobe. Both secondary electron imaging (SEI) and backscattered electron imaging (BEI) techniques were used. Part of the SEM analysis was performed with both energy dispersive x-ray (EDX) spectroscopy and wavelength dispersive x-ray (WDX) spectroscopy analytical methods.

WDX analysis was used to provide additional information not obtainable from the EDX analysis, such as the valence band transitions of silicon in the alloy compared to that in pure reference Si standards (Ref. 15). Specifically, the chemical state of silicon in the Mo-Ni + SiC alloy was determined together with the carbon content in different phases identified in the alloy. The Si Kα and Si Kβ analysis was performed using a polyethylene terephthalate (PET) crystal at 10 kV and with a beam current of 200 nA.

The C Kα analysis was performed using a LDEC crystal. Pure standards of silicon, silicon carbide, and carbon were used for calibration purposes. The average of a two-point background was subtracted from the measured peak intensities, and the positions were counted for 10 s. Estimates of the C K-ratios were determined sim-
ply from the following ratio: wt-% C = \( \frac{I_{unk}}{I_{std}} \times 100 \), where \( I_{unk} \) is the intensity of the unknown, and \( I_{std} \) is the intensity of pure C. The values reported in this study are not absolute; they represent ratios of the unknown to the standard.

Results

Joint Shear Strength

The load/displacement curves of all brazed and virgin Mo specimens upon testing are shown in Fig. 2. As seen in Fig. 2, all doped specimens are stronger than the undoped one. The best result, shear strength increase of about 26%, was obtained in the brazement with 3% SiC. The nonlinear shear strength concentration dependence with maximum value at 3% SiC doping, we hypothesize, was because this alloy had the finest mixture of the joint eutectic phases and advantageous morphology in the joint interface zone (see later for the justification). The failure of all brazed specimens occurred in the brazement zone and proceeded in a brittle mode. No plastic deformation stage is present on the loading curves in Fig. 2.

In the case of the molybdenum Z-specimen, failure occurred at about 436 MPa, which is close to the data on upper shear strength under torsion testing (Ref. 16). An agreement of the results for pure Mo, strain rate 1.25 mm/min, with that of the ASM standard data proves that the implemented measurements method is reliable to apply in brazement testing for characterizing the joint strength. The pure Mo specimen failure was also initiated in the ends of the overlap neck area (i.e., simulating brazement), with failure cracks propagating in the specimen body at about 45 deg upon failure. The strength of brazements manufactured in the current work is equal to, or even exceeds, 50% of that of pure Mo (i.e., twice as high as that obtained in Ref. 11).

Joint Phase Microhardness

The relative nano hardness of brazement phases was assessed via indent depth. The smaller the imprint, the higher the nano hardness. It can be ranked in the following order (from soft to hard): [Mo crucible, 5.5 nm] < [control 4-2A (0% SiC – Ni-rich phase, 6.8 nm] < [Control 4-2A – Mo-rich phase, 5.1 nm] < [5% SiC 9-2A – Ni-rich phase, 4.0 nm] < [5% SiC 9-2A – Mo-rich phase, 2.0 nm].

The AFM nano hardness analysis correlates with the EDX and WDX microprobe analysis results on silicon and carbon concentration in joint phases (Ref. 4). Both of these analyses revealed that silicon and carbon are dissolved as solutes in basic Mo-Ni and Ni(Mo) joint phases without changes in their crystal structure. It was also found that silicon and carbon concentrations are higher in the Mo-rich phases in the samples with nano SiC additions.

The observed increase of hardness
in both Ni- and Mo-rich phases was consistent with the increase of carbon and silicon concentrations. The strengthening effect relative to the nondoped sample was the most pronounced (about three times the increase) in the Mo-based phase of the alloy with 5% SiC. It can be assigned to the solid-solution strengthening effect in both phases (Ref. 17).

**Joint Microstructure**

Figure 3 represents joint microstructures of the interface domain for Mo/binary Mo,Ni specimens with 0, 1, 3, and 5 wt-% of SiC additions (see Table 1 for the samples’ designations and compositions). In all cases, alloys reacted well with pure Mo, forming a good joint.

For example, in the case of 0 wt-% (Sample material 4-2a, Table 1)(Fig. 3A), the interface between Mo and Mo-Ni, was well developed, showing a strong bonding of the crystallized Mo-

![Fig. 4 — A — Microstructure of Mo-Ni 4-2a (Table 1) control specimen without SiC (see Tables 2 and 3 for the phase composition at the locations A and B); B — microstructure of Mo-Ni specimen containing 1 wt-% SiC (7-2a). Note that there are two types of Mo-rich phases containing different concentrations of Mo and having different morphology; C — Microstructure of Mo-rich phase observed in Mo-Ni 7-2a specimen containing 1 wt.-% SiC; D — EDX spot composition analysis of Mo-rich phase observed in a specimen containing 1 wt.-% of SiC.](image-url)

**Table 2 — Phase Composition at Location A (Fig. 4A)**

<table>
<thead>
<tr>
<th>Elements</th>
<th>Wt-%</th>
<th>At.-%</th>
<th>K_A</th>
<th>K_F</th>
<th>K_Z</th>
<th>Intensity</th>
<th>P/bk g</th>
</tr>
</thead>
<tbody>
<tr>
<td>MoL</td>
<td>65.09</td>
<td>53.29</td>
<td>0.916</td>
<td>1</td>
<td>0.955</td>
<td>3037.409</td>
<td>0</td>
</tr>
<tr>
<td>NiK</td>
<td>34.91</td>
<td>46.71</td>
<td>0.974</td>
<td>1</td>
<td>1.079</td>
<td>1945.676</td>
<td>0</td>
</tr>
</tbody>
</table>

![Fig. 4 — A — Microstructure of Mo-Ni 4-2a (Table 1) control specimen without SiC (see Tables 2 and 3 for the phase composition at the locations A and B); B — microstructure of Mo-Ni specimen containing 1 wt-% SiC (7-2a). Note that there are two types of Mo-rich phases containing different concentrations of Mo and having different morphology; C — Microstructure of Mo-rich phase observed in Mo-Ni 7-2a specimen containing 1 wt.-% SiC; D — EDX spot composition analysis of Mo-rich phase observed in a specimen containing 1 wt.-% of SiC.](image-url)
Ni with Mo. No apparent defects at the interface (e.g., pores, etc.) are seen (Fig. 3A, inset). Joint crystallizations started at the interface with formation of a thin layer of the high-melting phase, namely Mo$_{0.34}$Ni$_{0.65}$ (Fig. 3A). This continuous, 5-10-μm thick layer consisted of rounded crystals of Mo-rich phase. The area in the vicinity of the Mo base metal surface, outlined by a dashed line in Fig. 4A, is enriched in Mo as a result of its dissolution in the liquid filler metal, and not sufficient compositional equilibration of the filler metal liquid phase during brazing (Tables 2 and 3). The excessive, relative to the pure eutectic composition, amount of Mo was present in Mo-Ni dendrites formed in the interface area.

In addition, the excessive Mo amount yielded formation of large dendrites of hypoeutectic NiMo crystals grown in the eutectic mixture, which had the equilibrium eutectic composition (Fig. 4A, circle A). Because the average alloy composition is close to the eutectic one, no Mo-Ni dendrite should form. Thus, the presence of these hypoeutectic crystals confirmed nonequilibrium distribution of Mo and Ni in the joint interface vicinity.

A more or less uniformed typical eutectic mixture was formed in the rest of the specimen (Fig. 4A, circle B). All crystals formed in this specimen Mo-Li phase, Band #3

Fine Mo/Si phases in bulk: Band #4

Intermediate Mo/Si phases near Interface: Band #3

Fine Mo/Si phases: Band #2

Large Mo/Si phases: Band #1

Mo Substrate

Table 3 — Phase Composition at Location B (Fig. 4A)

<table>
<thead>
<tr>
<th>Elements</th>
<th>Wt-%</th>
<th>At-%</th>
<th>K_A</th>
<th>K_F</th>
<th>K_Z</th>
<th>Intensity</th>
<th>P/bk g</th>
</tr>
</thead>
<tbody>
<tr>
<td>MoL</td>
<td>45.16</td>
<td>33.51</td>
<td>0.872</td>
<td>1</td>
<td>0.932</td>
<td>1863.818</td>
<td>0</td>
</tr>
<tr>
<td>NiK</td>
<td>54.84</td>
<td>66.49</td>
<td>0.982</td>
<td>1</td>
<td>1.053</td>
<td>2862.735</td>
<td>0</td>
</tr>
</tbody>
</table>

Table 4 — SEM/EDX Composition of the Mo-Rich Phase (Fig. 4B, point 1)

<table>
<thead>
<tr>
<th>Elements</th>
<th>Wt-%</th>
<th>At-%</th>
<th>K_A</th>
<th>K_F</th>
<th>K_Z</th>
<th>Intensity</th>
<th>P/bk g</th>
</tr>
</thead>
<tbody>
<tr>
<td>SiK</td>
<td>2.11</td>
<td>5.81</td>
<td>0.588</td>
<td>1.016</td>
<td>1.165</td>
<td>377.906</td>
<td>0</td>
</tr>
<tr>
<td>MoL</td>
<td>67.71</td>
<td>39.72</td>
<td>0.919</td>
<td>1</td>
<td>0.955</td>
<td>3188.172</td>
<td>0</td>
</tr>
<tr>
<td>NiK</td>
<td>30.17</td>
<td>39.69</td>
<td>0.972</td>
<td>1</td>
<td>1.079</td>
<td>1690.323</td>
<td>0</td>
</tr>
</tbody>
</table>

had rounded morphology. No faceted crystals typical for Mo-Ni intermetallic crystals and Mo-based carbides and silicides were observed. In the area 200–300 μm from the joint interface, the microstructure consisted of only a very fine mixture of Mo-Ni and Ni(Mo) eutectic phases. Their composition was close to an eutectic one, which can be found in the Mo-Ni phase equilibrium diagram.

A SEM image of the Mo-Ni filler metal + 1.22 at.-% SiC joint (Sample material 7-2a, 1 wt-% SiC) is given in Fig. 4B. A substantial number of well-faceted, Mo-enriched crystals were present in the eutectic crystal mixture (Fig. 4B, point 1, and Fig. 4C, D) beyond the interface area. Optical analysis revealed that these large, dozens-micron-sized Mo-rich crystals appeared to be much harder than the other phase structures based on variations in polishing edge retentions (see inset in Fig. 4C). Their composition (Table 4) was very close to that of the major Mo-Ni intermetallic phase existing in the Mo-Ni alloy system according to the Mo-Ni binary phase diagram (Ref. 18). They also contained 5.8 at.-% Si. This silicon concentration was higher than the average concentration of Si in the specimen. Therefore, it is obvious that Si preferred associating with the Mo-rich phase.
A large fraction of Mo-Ni crystals was present in this sample. These crystals did not have strong faceting (Fig. 4B inset, points 2 and 3). These areas were similar to eutectic areas in nondoped alloys. They also contained less of both Mo and Si. Still, variations of Mo, Ni, and Si concentrations in various MoNi crystals observed in this alloy were insubstantial. Some Mo-NiSi crystals had well-distinguished small spherical particles located in their centers.

The 3 wt-% SiC sample (Mo-Ni filler metal + 3.48 at.-% SiC joint, Sample material 8-3a) had a substantially different microstructure compared to the 1 wt-% sample. Its SEM micrograph is given in Fig. 5A. As seen in Fig. 5A, a narrow, Mo-enriched layer is attached to Mo in a similar way as in a sample with 0 and 1 wt-% SiC (arrows). The interface region between the Mo crucible and the bulk of the brazing filler metal has a number of bands of Mo- and Ni-rich phases — Fig. 5A. These bands are alternating layers of large, fine, and intermediate sized Mo/Ni phases. The bulk of the joint material has a relatively fine Mo-Ni phase located in the Ni-rich matrix Fig. 5A. EDX analysis clearly indicates that the Si content is much higher in the bright Mo-rich phases.

An interesting result of the study is that the Ni is higher in the large Mo-rich phases in Band 1 adjacent to the Mo interface, compared to the finer Mo-rich structures in Band 2. Also, Si was lower in the large phase structures in Band 1, adjacent to the Mo crucible. Again, the concentration differences were not great enough to pay special attention to these variations.

The microstructure of the Mo-Ni filler metal + 5.44 at.-% SiC joint (Sample material 8-3a, 5 wt-% SiC) shown in Fig. 5B consists of two distinct zones. The microstructure of Mo-Ni filler metal + 5.44 at.-% SiC joint (Sample material 8-3a, 5 wt-% SiC) shown in Fig. 5B consists of two distinct zones. First is the narrow interface zone seen in a SEM image and presented in Fig. 5B. This zone is located between the Mo base metal and the white dash outline, and it contains Bands 1, 2, and 3 — Fig. 5B. Second is the wide band of Mo-enriched crystals (point 2) separating the interface zone from the rest of the joints and is attached to the latter. The rest of the joint microstructure outside of the dashed line (Fig. 5B) consists of a fine, uniform mixture of two Mo-rich (point 3) and Ni-based (point 4) phases, which constitute the bulk of the eutectic. The phase dimensions in this area are smaller than 10 μm. Neither large Mo-Ni crystals, similar to those observed in the 1 wt-% SiC specimen, nor carbide or silicide crystals are found there.

As follows from the interface observations depicted in Fig. 5B, the interface zone is more or less similar in appearance, phase locations, and composition to that of the specimen containing only 1 wt-% SiC. Indeed, a narrow Mo-
enriched layer (arrows) is attached to Mo, followed by thin Ni-rich (point 1).

It is important to underline a drastic difference in morphology of the eutectic in the bulk of the joints having 0 vs. 3 and 5 wt-% Si. In the first case, the microstructure was a classical eutectic build of dendrite-like networks with rounded shapes. In joints containing 3 and 5 wt-% of SiC, it was an assembly of very fine cuboidal and trapezoidal particles of Mo-rich phase segregated in the Ni-rich matrix. All these particles had flat facets outlined by straight lines. The particle sizes were in order of a few micrometers. On the other hand, the morphology of the alloy with 1 wt-% SiC was transitional. It was a mixture of the classical, plate-like eutectic structure observed in the binary Mo-Ni alloy and newly observed combination of faceted crystals of the Mo-rich phase distributed evenly in the Ni-rich matrix. All these particles had flat facets outlined by straight lines. The particle sizes were in order of a few micrometers. On the other hand, the morphology of the alloy with 1 wt-% SiC was transitional. It was a mixture of the classical, plate-like eutectic structure observed in the binary Mo-Ni alloy and newly observed combination of faceted crystals of the Mo-rich phase distributed evenly in the Ni-rich matrix in the alloy with 5 wt-% SiC.

Figures 4 and 5 clearly demonstrate an evolution of joint microstructural changes in the Mo/Mo-Ni + SiC braze-metals due to an increase in SiC doping.

State of Silicon and Carbon in Joint Phases

To prove that Si and C exist only as solid-solution components, the WDX analysis was used as an independent analytical method. Preliminary EDX analysis was performed on a lighter-contrasting Mo-rich phase, revealing that this phase contained the highest Si content (Fig. 5B, point 2). WDX Si analysis was performed on Sample 9-2a, which nominally had the highest level, 10.88 at.-% (5 wt-%) of SiC addition.

WDX analysis can probe valence band transitions of silicon in the alloy, and compare those to pure silicon, and SiC reference spectra. A comparison of the spectra in Fig. 6 clearly shows the disappearance of Si-C peak in this specimen. This proved that Si-C bonding was absent in the Mo-Ni alloy with 5 wt-% SiC. There is also a small, but systematic, shift in the wavelength position between elemental Si, SiC, and Si in the alloy.

The WDX results prove that the Si and C were not in the form of SiC. Dissolution reaction of the SiC in the liquid Mo-NiSiC phase appeared to be complete. Both of these metalloids continued to exist as solid-solution components in both solid Mo- and Ni-rich phases upon crystallization. Again, it is obvious from the EDX and WDX data that Si was dissolved mostly in the Mo-Ni phase as a solid-solution component. The WDX data presented in Fig. 4 also showed that a small part of Si is dissolved in Ni(Mo) solid-solution phase. Indeed, the amount of Si in the Ni(Mo) solid solution is for an order of magnitude smaller (0.32 vs. 2.31%) than that in the Mo-rich phase.

In addition, a separate silicon profiling was made across the complete interface area, including a few dozen microns deep in Mo to determine both the silicon distribution in each eutectic phase crystal and the depth of Si diffusion penetration in molybdenum. The average silicon concentration across the interface decreased until it became equal 0% at about a dozen micron within the molybdenum base metal — Fig. 7. At the same time, it varied strongly between the Mo-base phase crystals, where it was in the range of 4.4 to ~2.3 wt-%, and Ni-based crystals, where Si concentration is about 0.2–0.3 wt-%. The Mo, Ni, and Si profiles showed that Si associated mostly with Mo, not with Ni — Fig. 8.

Such data showed that no noticeable Si penetration in Mo substrate (i.e., the base metal) was observed. The Si and C contents were relatively higher in the large Mo-rich phase, which is also consistent with the found hardness increase due to higher Si concentration in the solid-solution phase in accordance with the pertinent phase diagrams data for concentration ranges associated with the studied alloys given in Ref. 18.

It is important to outline that Mo-rich phases in alloy containing 1 wt-% of SiC had substantial elemental concentration variations of all three components. These variations are much smaller in alloys containing 3 and 5 wt-% of SiC. Within the middle of the joint with 5 wt-% of SiC, the distribution of silicon was more or less uniform among Mo- and Ni-based crystal.
The latter phenomenon is always an ex-
crystal rounding in lamellae formation.

12). The decrease in surface tension, in
transition metal surface tensions (Ref.
w28), well-known effect of Si additions on
added. This is the consequence of the
change of liquid and solid phases when Si is
added in Mo-Ni alloy. No evidence of Si
additions causes changes in the microstruc-
ture morphology of Mo-Ni eutectic braz-
ing alloy. The DTA data during
crystallization proved this hypothesis.

SiC should dissolve in liquid Mo-Ni
alloys with the formation of cuboidal and
trapezoidal eutectic crystals. The DTA data
during crystallization proved this hypo-
thesis. SiC addition to Mo-Ni alloy, the
Al addition to Mo-Ni alloy, the
larger the number of nucleating eutec-
tic nuclei. This dependence is much stron-
ger — N=exp(-A/\(\Delta T\))\(\gamma_{liq}\), where N is
the number of critical-sized nuclei form-
ing in the volume unit; A is a complex
constant that is weakly dependent on
composition; and \(\Delta T\) is temperature
supercooling (Ref. 20).

For all our samples, undoped and
doped, the supercooling was the same
during crystallization. It is no wonder that
Si, which has \(\gamma_{liq}=865\) dyn/cm vs.
\(\gamma_{liq}=1778\) for Ni and \(\gamma_{liq}=2250\) for Mo
(Ref. 12), refines eutectic microstruc-
ture due to decreasing the nuclei size
and increasing the crystallization rate. As
our observations demonstrate, the
more Si is added to Mo-Ni alloy, the
larger the number of nucleating eutec-
tic crystals and the smaller their size.

Conclusions

The shear strength of the brazed Mo-
Mo joint with the Mo-Ni eutectic
SiC doped filler was equal to, or even
exceeded, 50% of that of pure Mo. It
was twice as high as that reported for
undoped filler. The major beneficial
changes of the brazed joint’s mor-
phology due to alloying of Mo + Ni
alloys with SiC moderate additions take
place in the joint microstructure. These
changes are expressed in a transfor-
mation, a replacement of typical eutec-
tic morphology consisting of dendrites
imbedded within a mixture of lamellae
into a new type of refined microstruc-
ture consisting of a mixture of very
small, evenly distributed eutectic crys-
tals having cuboidal and trapezoidal
forms. This transition expressed itself
with changes in the forms and dimen-
sions of crystallized eutectic phase
components from rounded lamellae
crystals to faceted ones.

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Some parts of this work were reported
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WELDING RESEARCH

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**Appendix**

An additional analysis was made to clarify, semiquantitatively, the difference in Si concentration between Mo- and Ni-rich phases obtained via both EDX and WDX methods. For this, we used the Si K-ratio values representing ratios of Si in the unknown sample to that of the pure Si standard (Table A-1). Similar observations were performed for carbon, results of which are presented in Table A-2.

### Table A-1 — Silicon K-Ratio in Mo-Ni-Based Nano Composite Fillers According to SEM/WDX Analysis

<table>
<thead>
<tr>
<th>Sample</th>
<th>Area</th>
<th>Si K-ratio (%)</th>
</tr>
</thead>
<tbody>
<tr>
<td>7-2a 1 wt-% SiC</td>
<td>Areas in Sample 7-2a, Wt-% SiC</td>
<td></td>
</tr>
<tr>
<td>Large Mo phase</td>
<td></td>
<td>2.11</td>
</tr>
<tr>
<td>Large Mo particle, point 1</td>
<td></td>
<td>1.56</td>
</tr>
<tr>
<td>Ni-based matrix, point 2</td>
<td></td>
<td>0.09</td>
</tr>
<tr>
<td>Mo phase near interface</td>
<td></td>
<td>1.47</td>
</tr>
</tbody>
</table>

### Table A-2 — Carbon K-Ratio in Various Phases of Mo-Ni-based Nano Composite Fillers According to SEM/WDX Carbon Analysis

<table>
<thead>
<tr>
<th>Sample</th>
<th>Area</th>
<th>C Kα peak intensity (cps)</th>
<th>C Kα-ratio (%)</th>
</tr>
</thead>
<tbody>
<tr>
<td>Areas in Sample 7-2a, 1 wt-% SiC</td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>Large Mo-rich phase, Particle 1</td>
<td></td>
<td>625</td>
<td>1.10</td>
</tr>
<tr>
<td>Small Ni-rich phase, Particle 2</td>
<td></td>
<td>183</td>
<td>0.32</td>
</tr>
<tr>
<td>Small Ni-rich phase, Particle 3 in matrix</td>
<td></td>
<td>205</td>
<td>0.36</td>
</tr>
<tr>
<td>Areas in Sample 9-2a, 5 wt-% SiC</td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>C Kα peak intensity (cps)</td>
<td></td>
<td>206</td>
<td>1.24</td>
</tr>
<tr>
<td>9-2a 5 wt-% SiC in Mo-rich phase, point 1</td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>9-2a 5 wt-% SiC in Ni-rich phase, point 2</td>
<td></td>
<td></td>
<td></td>
</tr>
</tbody>
</table>

20-s WELDING JOURNAL / JANUARY 2017, VOL. 96
Introduction

Products are typically assembled using multiple components and joining processes essential to increase manufacturing process efficiency (Ref. 1). However, these joints are prone to frequent failures (Refs. 2–14). The causes and effects of dissimilar weld failures are listed in Table 1.

Figure 1 shows examples of various dissimilar weld (DW) failures. These failures can be examined by two aspects, microscopic and macroscopic. The microstructural aspects of the failures are shown in Fig. 1A and B. Figure 1A shows the typical morphology observed for welds made with Ni-based filler metals where a creep cavitation is initiated along the row of Type I carbides that form along the interface (Ref. 15). Figure 1B shows creep cavitation associated with the prior austenite grain boundaries (PAGBs) that occurs in welds made with stainless steel filler metals. From the macrostructural view, the majority of failures in ferritic-austenitic DW occur at the weld interface on the ferritic side of the joint, which is due to stress concentration — Fig. 1C and D. An example of oxide notching that can occur on the ferritic side is shown in Fig. 1D. The presence of tensile residual stress (RS) can promote cracking in welds and within the heat-affected zone (HAZ) during service in power plants (Ref. 5). These stresses cannot be reduced by heat treatment to the same extent as joints made of similar welds (Ref. 7). This is because the thermal and mechanical properties of two base metals (BMs) are not similar; for example, coefficients of thermal expansion (CTE) of ferritic steels and austenitic stainless steels are $11 \times 10^{-6}$ 1/°C and $20 \times 10^{-6}$ 1/°C, respectively (Refs. 16, 17).

During the last three decades, numerical simulations have been employed to investigate RS in welds and their evolution during welding (Ref. 18). For example, Deng (Ref. 17) predicted the RS produced in gas tungsten arc welding (GTAW) similar 1020 and 1045 steels using the finite element (FE) method. The result showed that the phase transformation does have an important influence on the RS distribution in 1045 steel but does not have a significant effect on 1020 steel.

However, there are three major challenges to produce a successful model of RS in DWs. First, the weld zone (WZ) in a DW is asymmetric, which is attributed to asymmetric flow field and velocity in the weld pool (WP) and thermal fields of BMs (Ref. 19). Typically, investigators assume a symmetric power density distribution in their RS analysis, even in DW (Refs. 8, 20–22, 32). These assumptions are introduced to cope with the complexity of the heat transfer analysis. However, these assumptions utilized in the model will not generate the asymmet-

Numerical and Experimental Investigation of Residual Stress Distribution in a Dissimilar Ferritic-Austenitic Weld

The martensitic phase has a significant influence on the transverse and longitudinal residual stress components

BY H. EISAZADEH, J. BUNN, AND D. K. AIDUN

ABSTRACT

In this study, a model considering an asymmetric power heat distribution, temperature-dependent material properties, strain hardening, and phase transformation was developed to predict the temperature field and residual stress distribution in a gas tungsten arc (GTA) dissimilar weld between austenitic stainless steel (AISI 304) and low-carbon steel (AISI 1018). The effect of martensite formation on the longitudinal and transverse residual stress distributions were investigated using both finite element (FE) model and neutron diffraction measurements. The results indicated that the martensitic phase had a significant influence on the transverse and longitudinal residual stress components. The martensitic phase does not only change the distribution of residual stresses near the weld centerline but can also alter the peak value of the residual stress. The calculated temperature and weld zone (WZ) profile were in agreement with the experimental results from thermocouples and a macrograph of the weld. Favorable agreement was also found between the calculated residual stress distribution from the FE model and residual stress measurements obtained by neutron diffraction.

KEYWORDS

• Finite Element (FE) Modeling of Residual Stress • Dissimilar Weld
• Martensite Phase • Neutron Diffraction
ric weld shape that forms in actual DW. In the absence of fluid flow analysis in RS studies, as suggested by Goldak et al. (Ref. 23), the double ellipsoidal power density distribution should be defined in a way that could result in an asymmetric weld shape.

Second, the effect of high-temperature material properties of the WZ on the numerical model should also be taken into consideration because its composition is different from the BM’s, and likewise its mechanical properties. For example, studies conducted by Lee et al. (Ref. 20), Javadi et al. (Ref. 21), and Ranjbarnodeh et al. (Ref. 22) have assigned mechanical properties of BMs to their corresponding side of the dissimilar WZ. This assumption can be valid as long as the size of the WZ is relatively small compared to the size of the weldment.

Third, the thermal history, especially during the cooling process, can significantly affect the overall RS distribution. Unfortunately, metallurgical changes cannot be observed directly and are problematic to mathematical modeling (Ref. 24). Although numerical models exist to predict the evolution of the metallurgical phases, accounting for volumetric changes, it remains a matter of debate as to the magnitude of the impact of the phase transformations on RS.

In this study, the residual stresses (RSs) induced by autogenous GTAW were numerically investigated. An asymmetric power density distribution model was used to match the data from experiments for the WZ geometry. The material properties of the WZ were updated as the heat source moved along the weld path. Both transverse and longitudinal RSs were captured and compared with the experimental results published in a previous study (Ref. 25). Outcomes of this work should improve the understanding of the role of phase transformation on the magnitude and distribution of RSs in the DW. In addition, to provide a more accurate history of RS formation and distribution, the mechanical properties of the WZ were updated during GTAW.

Experimental Procedures

Setup

Experiments were carried out using a
single-pass, autogenous GTA butt joint weld on AISI 1018 to AISI 304. Compositions of these metals are shown (Table 2). The welding parameters (Table 3) were designed so as to produce partial penetration because no auxiliary argon gas was accommodated for the weld protection underneath. After welding, the specimens were allowed to cool for 2000 s, after which the temperature had approximately equilibrated, before being released from the fixture. Dimensions of the samples are shown in Fig. 2.

Microstructure of the Dissimilar Weld Zone

The Schaeffler diagram was used to predict the microstructure of the WZ for a dissimilar weld. To calculate dilution, cross-sectional areas of the melted 304 and 1018 BMs were calculated with the aid of ImageJ image analysis software (Ref. 26). Calculation showed approximately 43% dilution from 1018 and 56% dilution from 304 BM. Considering the composition of the BM, \( C_{req} \) and \( N_{eq} \) were evaluated using Equations 1 and 2. Three points, representing the microstructures for the 1018, 304, and dilution of dissimilar weld, are shown in Fig. 3.

According to EDAX analysis, the result of which is shown in Table 2, \( C_{req} \) and \( N_{eq} \) were calculated. These results are plotted in Fig. 3, which are in good agreement with the measured dilution. According to the Schaeffler diagram, dissimilar WZ is martensitic.

\[
C_{req} = \%Cr + \%Mo + 1.5(\%Si) + 0.5(\%Nb) \quad (1)
\]

\[
N_{eq} = \%Ni + 30(\%C) + 0.5(\%Mn) \quad (2)
\]

Measurement of Residual Stresses

In this study, the neutron diffraction (ND) method was used to nondestructively measure the residual elastic stresses (Ref. 25). These measurements were performed at the High Flux Isotope Reactor of Oak Ridge National Laboratory (HFIR) on the 2nd Generation Neutron Residual Stress Facility (NRSF2). In the experiment, two different gauge volume geometries were utilized. The gauge volume

---

**Table 2 — Chemical Composition of the Plates and Weld Zone (wt-%)**

<table>
<thead>
<tr>
<th>Composition</th>
<th>C</th>
<th>Mn</th>
<th>Si</th>
<th>P</th>
<th>S</th>
<th>Cr</th>
<th>Ni</th>
<th>Balance</th>
<th>( C_{req} )</th>
<th>( N_{eq} )</th>
</tr>
</thead>
<tbody>
<tr>
<td>1018 steel (average)</td>
<td>0.16–0.20</td>
<td>0.30–0.90</td>
<td>0.15–0.30</td>
<td>0.04</td>
<td>0.05</td>
<td>—</td>
<td>—</td>
<td>Fe</td>
<td>5.7</td>
<td>0.3</td>
</tr>
<tr>
<td>304 stainless (average)</td>
<td>0.08</td>
<td>1.7</td>
<td>0.52</td>
<td>0.045</td>
<td>0.035</td>
<td>18.9</td>
<td>7.5</td>
<td>Fe</td>
<td>9.6</td>
<td>19.6</td>
</tr>
<tr>
<td>1018-304 WZ (EDAX)</td>
<td>—</td>
<td>1.14</td>
<td>0.12</td>
<td>—</td>
<td>—</td>
<td>7.58</td>
<td>4.32</td>
<td>Fe</td>
<td>6.7</td>
<td>8.3</td>
</tr>
</tbody>
</table>

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**Table 3 — Welding Parameters Used in This Study**

<table>
<thead>
<tr>
<th>Arc Voltage (V)</th>
<th>Arc Current, DCEN (A)</th>
<th>Electrode Diameter (mm)</th>
<th>Arc Length (mm)</th>
<th>Travel Speed (mm/s)</th>
<th>Argon Gas (m³/min)</th>
<th>Arc Efficiency</th>
</tr>
</thead>
<tbody>
<tr>
<td>18</td>
<td>150</td>
<td>2.4</td>
<td>1.5</td>
<td>4</td>
<td>0.0136</td>
<td>80% (Ref. 19)</td>
</tr>
</tbody>
</table>
where $d$ and $d_i$ are the interplanar spacing under the stressed and stress-free state, respectively. And $\theta$ and $\theta_i$ are the diffraction angles, in radians, for the stressed and stress-free specimens at each location, respectively (Ref. 27).

$$
\varepsilon_i = (d - d_i)/d_i = -\cot\theta_i (\theta - \theta_i) \quad (3)
$$

where $d_i$ is required as a reference in the ND strain calculation, a macroscopic “stress-free” comb-like coupon was used. Each comb-like coupon was cut from a nominally identical weld at the same locations as the ND measurements using the electric discharge machining (EDM) process — Fig. 4. The coupon has been sliced 3 mm wide and 3 mm thick along the transverse direction to effectively release the macroscopic stresses from the bulk of the weldment.

By assuming that the three orthogonal components of the measured strain correspond to principal directions, meaning that shear strains were assumed zero in these defined directions, the analysis to determine RS was simplified. The macroscopic stress components, where the numerical superscripts refer to the lattice plane family, were related to the elastic strains in analogy with Hooke’s Law by (Ref. 27):

$$
\sigma_{ij} = E_{ijkl} / (1 + \nu_{ijkl}) [\varepsilon_{ij} + \delta_{ij} \nu_{ijkl} (1 - 2\nu_{ijkl}) \varepsilon_{kl}] \quad (4)
$$

where $E_{ijkl}$ and $\nu_{ijkl}$ are the diffraction elastic constants relating strain in the (311) and (211) lattice planes to the macroscopic stress for the FCC and BCC phases. In addition, $\varepsilon_{ij}$ and $\varepsilon_{kl}$ in the above equations are the elastic bulk strain, which can be found in the table of constants (Ref. 25). $\delta_{ij}$ is the Kronecker delta function.

Finite Element Modeling

The welding process simulation consists of a thermal analysis in which the temperature and martensite phase evolution in the WZ are determined as a function of time, followed by a mechanical analysis, which employs the temperature history obtained from the thermal analysis. The dimensional changes in arc welding are negligible, and mechanical work done is insignificant compared to the thermal energy generated by the arc; therefore, sequentially coupled analysis (SCA) works very well (Ref. 28). In SCA, the formulation considers the contributions of the transient temperature field to the stress analysis through thermal expansion, as well as temperature-dependent mechanical properties. A benefit of this SCA was saving a large amount of computational time. A fine mesh was used for the high-temperature gradients in the weld and its vicinity ($\pm 6$ mm from the weld centerline), and the element sizes gradually increased with the increase of distance from the weld centerline — Fig. 5.

Thermal Analysis

A full 3D model with 27000 8-node linear heat transfer brick elements (DC3D8) was built using the ABAQUS FEM program (V6.11) (Ref. 28). The initial temperature of the material was set to 25°C. Thermal boundary conditions consist of the application of conductive, convective, and radiative heat transfer to all surfaces of the model, which were used as a function of the metal surface temperature. The metal surface in the WZ and the surrounding region, being at higher temperature, is dominated by heat losses due to radiation, while losses due to convection dominate in the relatively lower temperature surfaces, away from the WZ. The temperature-dependent convection coefficients were used in our study (Table 4) for steel surfaces (Ref. 39). A value of 800 W/m²K was considered for surfaces that were in contact with the fixture (Ref. 8).

To reproduce the asymmetric DW in experimental results, the currently
Mechanical Analysis

In this analysis, all brick elements used in the heat transfer model were substituted with 8-node linear structural brick elements (C3D8R). To prevent rigid body motion without constraining the deformation of the WZ, displacements of all nodes along the Y direction and selected nodes in Fig. 4 were restricted. Accurate constitutive material models were highly important for the prediction of stress and strain development. In the current simulation, the total strain ($\Delta\varepsilon_{\text{total}}$) induced by the welding process was divided into elastic ($\Delta\varepsilon^e$), thermal ($\Delta\varepsilon^t$), volume change due to martensite formation ($\Delta\varepsilon^m$), plastic ($\Delta\varepsilon^p$), and plastic due to phase transformation ($\Delta\varepsilon^pt$) — Equation 5. The effect of plastic strain due to phase transformation ($\Delta\varepsilon^pt$) was not taken into account in this study.

$$\Delta\varepsilon_{\text{total}} = \Delta\varepsilon^e + \Delta\varepsilon^t + \Delta\varepsilon^m + \Delta\varepsilon^p + \Delta\varepsilon^{pt} \quad (5)$$

The elastic behavior ($\Delta\varepsilon^e$) was modeled using isotropic Hooke’s law with temperature-dependent Young’s modulus and Poisson’s ratio (Ref. 28). The thermal strain ($\Delta\varepsilon^t$) was computed using the temperature-dependent CTE (Ref. 28). For the plastic behavior ($\Delta\varepsilon^p$), the von Mises yield surface criterion (Ref. 28) was considered — Equation 6 (Ref. 31).

$$f = \sigma - \sigma_y \geq 0 \quad (6)$$

where $\sigma$, $\sigma_y$, $\sigma$, and $\sigma_0$ are the initial yield stress, plastic strain, and material constants, respectively. Figure 7 gives an idea of the behavior of the isotropic hardening, in which the yield function is considered in two-dimensional principal stress space by putting $\sigma_0 = 0$. In this numerical simulation, a three-dimensional yield function is considered.

The effect of austenite-to-martensite transformation $\Delta\varepsilon^m$ has also been taken into consideration in the numerical model (Ref. 33). A general schematic of the volume change due to phase transformation for plain carbon steel is shown in Fig. 8A. During welding, when the steel reaches the temperature $A_1$, it begins to transform from pearlite-ferrite to austenite, and when the temperature reaches $A_3$, the austenite transformation is assumed to be complete and the material is fully austenitic. Ferritic steel has a BCC structure, whereas austenite has a FCC structure. During austenitic transform-

![Fig. 6 — The power density distribution model with four different octants for the dissimilar weld. The x axis represents the boundary between the two base metals.](image)

![Fig. 7 — Isotropic hardening, in which the yield surface expands with plastic deformation and the corresponding uniaxial stress-strain curve (Ref. 31).](image)
In this study, material dilution was used for all properties of the WZ except yield stress. This assumption may induce some uncertainty calculations, which was ±20 MPa for residual stress distribution. The yield stress of the WZ was assumed 550 MPa at room temperature (Refs. 34, 35). Zhu and Chao’s (Ref. 36) approach was used as an engineering approximation of the yield stress at high temperatures for the WZ, which is shown in Equation 12.

\[ \sigma_y = 550 \, \text{MPa} \quad 273 \leq T \leq 373^\circ K \\
\sigma_y = \frac{5\%\sigma_y + \frac{1100-T}{1000} \times 95\%\sigma_y}{373 < T < 1100^\circ K} \\
\sigma_y = 10 \, \text{MPa} \quad T \geq 1100^\circ K \]

where 1100°K is the cut-off temperature (i.e., about 2/3 of the WZ melting temperature of 1650°K), and 550 MPa is the WZ’s yield stress. The yield stress of the WZ at different temperatures is shown in Fig. 10E.

**Results and Discussion**

**Temperature Study**

To verify the accuracy of thermal analysis, the WZ profile predicted by the FE analysis was compared with the actual cross section of the weld. As shown in Fig. 11, the WZ of the AISI 304 side is deeper than that of the 1018 side because of different physical properties such as the lower conductivity and lower melting temperature.
of AISI 304. The profile of WZ predicted by FE modeling is about 27% less than the actual WZ. This is because the fluid flow and mass transfer in the weld pool were not considered in this study. Although authors are aware of other numerical models (Refs. 19, 37) that can produce an asymmetric dissimilar weld zone, these works run thermal and fluid analysis and did not include the mechanical analysis to predict stress distribution, because it makes the simulation complex, cumbersome, and time consuming. Therefore, among the FE models that predict temperature and stress simultaneously, current work has the newest power density model for dissimilar welds because, without performing fluid analysis, it is capable of capturing asymmetric shapes of dissimilar weld zones.

The FE model was also validated by comparing the predicted temperature histories with experimental data obtained from thermocouples. The temperature histories at six different locations (three on the AISI 304 side and three on the 1018 side) were measured using K-type thermocouples spot welded to the upper surface of the BMs. The thermocouples were placed at 10, 15, and 20 mm away from the weld centerline — see Fig. 5. And the temperature data was acquired at a rate of 10 Hz.

As shown in Fig. 12, the overall model prediction agrees well with the measurements. Near the peak, the model agreement was much better for locations 15 and 20 mm, when compared to location 10 mm. This could be attributed to the difference in the proximity of these locations to the WZ. Being the nearest to the WZ, the 10-mm point experiences the largest temperature change.

Comparison of the temperature histories shows that the peak temperatures in AISI 304 BM are higher than those of 1018 (especially for 10 and 15 mm), because the thermal conductivity of the former is much less than the latter. Another reason can be the lower specific heat of AISI 304.

Effect of Phase Transformation on the Longitudinal RS

First, the longitudinal RSs in the weldment are addressed because the stresses accumulated in the dissimilar weldment (DW) were the greatest in the longitudinal direction. Figure 13 shows the predicted 3D longitudinal RS field in the DW. Figure 13A, which does not include the effect of phase transformation, shows there exists a high RS concentrated near the weld centerline. This distribution is uniform through the thickness. In general, RS formed in the thin weld plate is
relatively uniform in the thickness direction because the thermal gradients are not significant. Another reason is that it was assumed the ferrite phase of the AISI 1018 plate and the austenite phase of the AISI 304 plate in case 1 do not undergo the phase transformation as the welding process carries on for case 1, neither in the WZ nor HAZ.

However, in the model considering the phase transformation, RS is no longer uniform across the thickness — Fig. 13B. As shown in the interface of AISI 1018 and 304, a gradient is now evident in the thickness direction as seen in Fig. 13B, indicating a low magnitude of tensile RS on the top surface, and a high tensile RS at the mid surface. The reason behind this change is attributed to the existence of the martensite phase in the latter case. During cooling, when the FCC structure of the WZ (austenite) transforms to the BCT structure (martensite), a volume expansion is undergone. This expansion produces a transformation-induced compressive stress component within the WZ; however, it induces a tensile stress to the surrounding area, according to the modeling results. This pronounced tensile RS region is observed near the mid surface of Fig. 13B. In particular, the compressive RS induced by phase transformation is balanced by the surrounding area (within ±5 mm). And the effect of the phase transformation is localized in the WZ.

Figure 14 shows the predicted longitudinal RS on the top and mid-surfaces of DW as well as the ND measurement on the mid thickness of the same weld. It shows all three numerical results followed a similar trend except the region close to WZ, which is within the range ±10 mm. Further, both case 1 and 2 (midsurface) were matched well, even within this range (±10 mm). The reason is that case 1 did not include any phase transformation effect and its typical model resulted in a belled-shape longitudinal RS. This shape was mentioned by several studies (Ref. 38). Similarly, the mid-surface of case 2 did not experience any phase transformation. However, its magnitude was somewhat increased by an amount of tensile RS caused by volume change due to martensite phase formation, while maintaining its belled-shaped trend.

The RS in the top surface of case 2 behaved in a different way when compared to its mid-surface. A large RS gradient exists in the center of the former because it contains the martensite phase while the latter does not. Comparison of the numerical results with ND measurements illustrates that the midsurface of case 2 matches best. The reason is because the ND measurement was fulfilled with the (211) plane.
on the AISI 1018 side and the (311) plane on the AISI 304 side, and its setup was not changed to capture RS in the martensite phase. Some deviation exists in the ND and numerical results. These could be due to the lack of enough accurate material properties data of the WZ in the numerical model, especially in high temperature. In ND measurement, they could also arise due to the assumptions of principal strains made when performing RS analysis utilizing 3 orthogonal strains obtained using ND, as it is reasonable to expect that the principal directions near the WZ may not lie directly in the transverse or longitudinal directions of the plate. Furthermore, moving a sample gauge volume across buried edges of phase transformations (such as those in the WZ) can be problematic. The analysis method utilized in the ND study attempts to correct for this, but some variations could still be present in the dataset as a result.

**Effect of Phase Transformation on Transverse RS**

Details of transverse RS in both similar and dissimilar welds (DW) are usually ignored because their magnitudes are much lower when compared to longitudinal RS (Refs. 7, 17, 20, 22). However, this work addresses transverse RS because of its significant impact on RS distribution in the WZ of a DW.

Figure 15 shows the predicted 3D transverse RS field in DW. Figure 15A, which does not include the effect of phase transformation, shows that the distribution of RS is uniform throughout the model excluding the small variation close to the weld centerline. However, in the model considering the WZ martensite phase transformation, the uniformity of RS no longer exists in the WZ — Fig. 15B.

As shown in Fig. 16, the transverse RS magnitude of both the model (case 2) and the ND results are smaller than the RS values in the longitudinal direction because the contraction due to material shrinkage in the longitudinal direction is greater than that of the transverse direction.

It is interesting to note that the RS in the transverse direction at the top (dashed blue) and mid (solid blue) surfaces of the case 2 model undergoes a sharp decrease and increase near the WZ (±30 mm), respectively. The reason for this drastic change is the existence of the martensite phase in the top surface of the WZ.

**Conclusions**

A 3D thermal analysis of GTAW AISI 1018 to 304 BMs with asymmetric power density distribution showed that the predicted WZ profile and thermal cycles agreed with the experimental results.

The peak temperature in AISI 304 was higher than that of the AISI 1018 side because AISI 304 has lower thermal conductivity.

The model presented has the capability to capture residual stress changes due to martensite transformation and update the WZ’s material properties as a result of weld dilution. Austenite to martensite transformation within the WZ not only changes the longitudinal RS but also the transverse RS.

The calculated longitudinal and transverse RSs were in good agreement with the ND measurements.

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**References**


**Fig. 15 — Transverse RSs (S33) distribution in the two cases: A — Without considering the phase transformation effect; B — with the phase transformation effect.**

**Fig. 16 — Transverse RSs (S33) along the ND measurement path as shown in Fig. 2.**


