



Review of Relationship Between Particle Deformation-Coating Microstructure and Properties in High Pressure Cold Spray

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Abstract

In the cold spray (CS) process, deposits are produced by depositing powder particles at high velocity onto a substrate. Powders deposited by CS do not undergo melting before or upon impacting the substrate. This feature makes CS suitable for deposition of a wide variety of materials, most commonly metallic alloys, but also ceramics and composites. During processing, the particles undergo severe plastic deformation and create a more mechanical and less metallurgical bond with the underlying material. The deformation behavior of an individual particle depends on multiple material and process parameters that are classified into three major groups - powder characteristics, geometric parameters, and processing parameters, each with their own subcategories. Changing any of these parameters leads to evolution of a different microstructure, and consequently change the mechanical properties in the deposit. While cold spray technology has matured during the last decade, the process is inherently complex and thus

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the effects of deposition parameters on particle deformation, deposit microstructure, and mechanical properties remain unclear. The purpose of this paper is to review the parameters that have been investigated up to now with an emphasis on the existent relationships between particle deformation behavior-microstructure-mechanical properties of various cold spray deposits.

Keywords: Cold spray, High pressure cold spray, Particle deformation, Microstructure evolution, Mechanical properties

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1. Introduction

1.1. Overview of cold spray

CS is a deposition/consolidation process in which powder particles (typically 5–50 µm) are accelerated to speeds of 300-1400 m/s by a high-pressure carrier gas as the gas expands in the divergent section of a DeLaval nozzle [1-5]. To achieve greater gas flow velocities in the nozzle, the compressed gas is often preheated and preheat temperatures as high as 1100°C are used, but are typically in the range of 40-70% of the particle melting temperature. Because the contact time of spray particles with the hot gas is quite short and the gas is rapidly cooling as it expands in the diverging section of the nozzle, the temperature of the particles remains substantially below the initial gas preheat temperature. Hence, the deposition can be easily controlled to occur below the melting temperature of the feedstock powder material, producing deposits from particles in the solid state. Because of this, CS deposits are essentially free of thermally induced defects commonly observed in traditional thermal spray deposits, such as oxidation, evaporation, gas release, shrinkage porosity, and thermally induced residual stresses [5-8]. Because of these advantages over other spray deposition processes, CS has generated a great deal of interest within the manufacturing community for repair, advanced coatings, and additive manufacturing applications. But, CS just like other deposition processes, has its own advantages and limitations, which are discussed below.

1.2. The advantages and limitations of CS

1.2.1. Advantages:

Thermal spray technology has been widely used in aerospace, defense, and gas turbine industries for fabrication of parts and components, preparation of protective surfaces, and refurbishment and repair of mismachined and service-damaged parts [2, 4, 5]. In recent years, CS has been introduced as a more reliable yet affordable process to produce various metal, ceramic, and composite depositions with superior qualities for these same purposes [5].



Fig. 1. Differences in particle velocity, carrier gas temperature, and substrate temperature between CS and thermal spray processes (Ref 9)

As shown in Fig. 1, CS differs from conventional thermal spray methods in the particle velocity, gas temperature, and substrate temperature. The advantages of CS over conventional thermal spray processes can be summarized as follows:

1- CS is well-suited for deposition of temperature-sensitive materials such as nanocrystalline (NC) and noncrystalline materials, oxygen-sensitive materials such as aluminum (Al), copper (Cu) and titanium (Ti), and phase-sensitive materials such as carbide composites, because of the low deposition temperature [9, 10].

2- CS of metals generally enhances fatigue resistance because of the micro "shot peening" effect, which generates compressive residual stresses in the resultant deposits [11, 12].

Method Porosity, %		Oxygen content, %		
CS	0.1 (Ref 14)	Same as in the powder (Ref 5)		
HVOF	1 (Ref 13)	0.2 (Ref 15)		
FLAME	10-20 (Ref 13)	4-6 (Ref 15)		
ARC	10-20 (Ref 13)	0.5-3 (Ref 15)		
PLASMA	1-7 (Ref 13)	0.5-1 (Ref 15)		

Table 1 Porosity and oxygen contents of deposits in different thermal spray methods (Ref 5, 13-15)





3- Metal CS deposits contain microstructures with high degrees of consolidation similar to wrought alloys due to the intrinsic high energy-low temperature features [16].

4- CS deposits feature higher thermal and electrical conductivities because of the higher density and reduced presence of oxide phases (Table 1) [5, 9, 17].

5- CS features significantly greater deposition efficiency (DE), as shown in Fig. 2 [5, 18], although equal or lower deposition rates.





7- CS generally increases the possibility of dissimilar materials joining because of less heat input into the substrate, which makes the substrate material to be of a less importance in this process [19, 20].

1.2.2. Limitations

As with any materials processing technique, the CS process has limitations. Here we list the main limitations and disadvantages, and briefly described them.





- 1- Materials with little capacity for low temperature deformation are not well-suited to cold spray [5, 21, 22]. Because CS is a solid state deposition process, only materials with sufficient ductility in the processing temperature window can be deposited by this process. In recent years, however, research efforts have attempted to expand the range of cold sprayable materials [23-25], and the use of high pressure cold spray (HPCS) systems instead of low pressure CS (LPCS) have enabled practitioners to partly overcome this limitation.
- 2- Particles usually experience high deformation during CS, causing a loss of ductility in the CS deposits [26]. However, this problem can be partly mitigated by using proper feedstock powder (proper size and temper), or performing thermal post-CS processing, as described in section 4.4.3.
- 3- Like other spray deposition processes, CS is a line-of-sight process [27], which makes it difficult to deposit onto internal surfaces. However, with a stand-off distance of 15 mm commonly used in CS, one can design nozzle assemblies to fit within cavities that would otherwise be impossible to spray. At the time of this write-up, there are some manufacturers of commercial CS equipment that are able to supply nozzle assemblies capable of spraying materials in inside diameters as small as 90 mm [9, 10].
- 4- The substrate material must be of sufficient hardness relative to the powder to induce substantial plastic deformation in the incoming particles and achieve satisfactory bonding [17].
- 5- Gas consumption in CS is much greater than in most thermal spray processes because of the high velocities and flows required to propel the particles. However, with newer gas recovery systems (developed by US Army Research Laboratory (ARL) [28]) used through the processing, over 80% helium (He) recovery is possible.





6- Standards for CS have not yet developed. To date, only one standard specification exists -MILSTD- 3021, developed by ARL, 2008 [29]. However, companies using CS employ internal specifications for cold spraying various materials, and these are likely to evolve into widespread standards.

1.3. Comparison of HPCS and LPCS

CS processes are generally divided into two categories – high pressure cold spray (HPCS) and low pressure cold spray (LPCS) – each have advantages and limitations, and should be viewed as complementary. In reality, pressure is simply a process variable in a continuum across a very large process window, but the distinction comes from common equipment design variations that arise as a result of dealing with only lower pressures, or primarily high pressures. The distinctions between the two CS processes can be summarized as follows:

1- In the HPCS process, the solid powder feedstock particles mix with the carrier gas in the pre-chamber zone. They are then fed into the gas stream, upstream of the converging section of the nozzle, as shown in Fig. 3(a). However, in LPCS process, powder particles are introduced downstream of the throat section of the nozzle, perpendicular to the diverging part of the nozzle, and accelerated toward the substrate, as shown in Fig. 3.

2- In LPCS, the carrier gas is heated only in the spray gun, while in the HPCS process, gas is preheated in a separate heating unit as well as in the spray gun (Fig. 3).

3- The amount of powder that can be fed per unit time increases with increased gas flows, and thus can be greater in the HPCS process [30]. Typical spraying parameters used in these two CS processes are summarized in Table 2. The main differences relate to process gas, pressure level, and electrical power used.

4- For HPCS, a high-pressure powder feeder operated at a pressure near or greater than the main gas stream must be used to allow mixing of the main gas and powder feed line. High pressure





powder feeders are also usually larger and more costly than the feeders needed for low pressure use.

5- A major challenge associated with all cold spray processes is nozzle clogging, which can become more severe as particle velocity and temperature are increased. To overcome the problem, a blended powder can be used, in which larger or harder particles are added to the first particle population as well as using various clogging resistant nozzle materials [5, 9, 17].



Fig. 3 Schematic presentation of CS processes, (a) HPCS (Ref 7), and (b) LPCS (Ref 30)

Spraying parameter	HPCS	LPCS	
Process gas	N2, He, mixture	Compressed air	
Pressure, bar	7-40	6-10	
Preheating temperature	RT-1000	RT-650	
Gas flow rate, m3/min	0.85-2.5 (N2), max. 4.2 (He)	0.3-0.4	
Powder feed rate, kg/h	4.5-13.5	0.3-3	
Spraying distance, mm	10-50	5-15	
Electrical power, kW	17-47	3.3	
Particle size, µm	5-50	5-30	

Table 2 Typical	spraving	parameters	used in	CS	processes
I able a I plea	Spreying	parameters	ubcu III	$ \omega$	p1 0 0 0 0 0 0 0 0





6- Another common issue for both processes is erosive wear of the nozzle throat. This wear can affect the nozzle operation and lead to variations in operating conditions and deposit quality. This problem becomes more severe when hard particles are sprayed if wear resistant materials such as sintered tungsten carbide are not used for the nozzle.

7- LPCS generally requires less costly equipment because the pressures are lower. However, the nozzle design in LPCS is restricted to a lower range of expansion Mach number (usually <3), and the inlet pressure is also restricted (normally 1.7 MPa) to ensure that atmospheric pressure is sufficient to supply powders to the nozzle. As a result, lower particle velocities can be reached through the downstream powder feeding technique.

8- The primary advantages of the HPCS over LPCS are the larger materials selection and better quality of the deposits, although the investment costs for HPCS equipment are greater than those for LPCS. In addition, the HPCS can be more efficient, with greater gas flow and powder feed rates.

1.4. Relationship between particle deformation, microstructure, bonding, & properties

Microstructural characterization and finite element analysis (FEA) show that the powder particles experience high strains (up to 10) and high strain rates (up to 10⁹/s) during CS [31-33], yielding densities close to those of bulk wrought materials. Studies have demonstrated that CS can be used to produce dense deposits of a wide range of materials, including aluminum (Al), copper (Cu), nickel (Ni) and its alloys, 316L, titanium (Ti), and Ti-6Al-4V, on dissimilar substrates, including glass [34] and polymers [35]. Metal matrix composites (MMCs) and free forms with custom-graded properties can also be deposited by CS [36, 37].

A much wider range of materials can be deposited by HPCS systems compared to LPCSs. This is because HPCS systems generate substantially higher particle impact velocities and thereby





increase the occurrence of high plastic deformation at interface with the substrate and with adjoining particles [38-40]. The formation of the deposit depends strongly on the high strain rate deformation via particle impacts, and thus the effects of material and process parameters on particle deformation are of critical importance. Here, we attempt to summarize the controlling parameters and their effects on particle deformation behavior.

In addition, HPCS systems generate microstructural changes within the deposition, particularly at the interfaces via particle-substrate and particle-particle interactions stemming from the high velocity particle impacts. Microscopy tools, including light microscopy (LM), scanning electron microscopy (SEM), electron back-scattered diffraction (EBSD), and transmission electron microscopy (TEM), have enabled researchers to characterize and analyze the microstructure of materials produced by HPCS. These efforts have led to reports on the associated microstructural phenomena, including ultra fine grained (UFG) structure [41-46], increases in dislocation density resulting from work-hardening [47-51], recovery (RV) and recrystallization [31, 32], in-situ or post-CS precipitation [41, 42, 52, 53], residual stresses [11, 12], phase transformation [54-56], and localized amorphization [57-60].

The microstructural phenomena that occur in CS deposits can result in significant changes and variations in the local mechanical properties at the micron and sub-micron level. These changes can have adverse or beneficial effects on the performance of the deposit. For example, dynamic recrystallization reportedly occurs at particle-particle and particle-substrate interfaces because of severe plastic deformations (SPD) during CS, producing UFG structures at these sites [43, 45] and local increases in hardness [32, 41]. Thus, to design and create a deposit with specific mechanical properties, we must understand the various impact-induced microstructural phenomena. In this paper, we review these changes and the relationships between process/material parameters on





particle deformation, microstructural evolution, and mechanical property variations that arise in CS deposits. Where possible, we include suggestions on how to utilize available mechanisms to obtain desired microstructures and mechanical properties. The review is intended as a reference for researchers studying phenomena occurring during HPCS and for industrial practitioners seeking to apply the process in engineering applications.

2. Particle deformation behavior

Particle deformation is an intrinsic aspect of CS deposition, affecting both particle and substrate bonding, surface topography, and metallurgical processes (such as work-hardening) within the deposit. Because deposit formation depends on plastic deformation via particle impacts, particle deformation behavior is an important factor in the during deposition. The extent of particle deformation depends on two major parameters during CS deposition - particle velocity and particle temperature [5, 18, 61, 62]. Parameters that affect particle velocity and temperature, in other words particle deformation, can be classified in three categories - 1- powder characteristics (size, shape and state, surface oxide layer), 2- geometric effects (spraying stand-off distance, incidence angle, and nozzle geometry), and 3- process parameters (gas type, temperature and pressure, substrate hardness, temperature, and surface roughness). Here, we attempt to summarize the key findings of published reports on this subject. This section is intended as a reference for researchers studying effects of parameters on particle deformation during HPCS.

2.1. Powder characteristics

2.1.1. Particle size

Particle size influences the critical velocity and thus the bonding during the CS process [3]. Following the deposition, the rate at which the material loses heat plays an important role in bonding. Beside the heat capacity of the spraying material as well as its thermal conductivity, the cooling rate of the material is affected





by particle size and it decreases with increasing particle size. The cooling rate should be "low enough" to promote shear instability and on the other hand, "high enough" to allow the interface to cool down quickly and complete the bonding process. Shear instability can be hindered in small particles due to high thermal gradients with their surroundings [63]. Moreover, smaller particles experience higher quench rates during deposition, which increases the strength in these particles. Also, smaller particles will generally exhibit greater levels of impurities after production due to greater surface-to-volume ratios. Because of these factors, an optimum particle diameter exists above which particles experience sufficient plastic deformation and quench rates. Schmidt et al. [3] proposed the following equation for the critical dimension of particles, above which particles will adhere to deposit:

$$d_{crit} = 36 \frac{\lambda_p}{c_p \cdot \rho_p \cdot V_p}$$
 Eq. 1

where λ_p is the thermal conductivity, c_p is the specific heat of the particle, ρ_p is the density of particle material, and V_p is particle velocity. Based on Eq. 1, the critical diameter of different materials can be calculated. Fig. 4 shows the measured values for different materials using the equation. These values indicate that for tin, copper, silver and gold, thermal diffusion limits bonding of small particles, whereas spraying of titanium and steel 316L is less restricted by this effect.





Fig. 4 Minimum particle diameter for adiabatic shear instability for different materials (Ref 3)

In the same study [3], the authors showed (Fig. 5) that for a given particle contact velocity, a larger particle causes a stretched time scale, delaying and extending the peak and post-peak temperature. The temporal extension of the hot region for larger particles was attributed to the increased times required for the deformation wave to reach and propagate through the monitored volume. The results reported also demonstrate that bonding quality will increase with particle size under the same impact velocities, because maximum temperature and the corresponding bonding time are both increased.



Fig. 5 Effect of Cu particle size on temporal evolution of the temperature and time scale (Ref 3)





2.1.2. Powder state and shape

How the feedstock powder is produced, e.g. by gas atomization or cryomilling, as well as the particle shape, e.g. spherical or irregular, strongly influence the particle deformation behavior during cold spraying and thus the quality of CS deposits [64-66]. Powders which may have the same chemical composition but different physical characteristics, will thus require different production methods [65], a factor that is critical for the design and optimization of processes and equipment.

In one study by Wong et al. [66], various Cu powder shapes were deposited on Cu substrates by CS to determine the effect of the feedstock powder shape on the deformation behavior and the deposition process. Each powder was sprayed under six gas pressure-gas temperature conditions, which are presented in Table 3. In all conditions, the largest average particle impact velocity was attained for irregular powder shapes followed in order by a sponge powder and finally, spherical powders (Fig. 6). For spherical powders, greater particle impact velocities were achieved with smaller particles. The non-spherical morphologies of irregular and sponge powders caused variable drag coefficients, which resulted in extensive variations in particle acceleration and particle impact velocity [67]. Consequently, the sponge and irregular powders had broader distributions of measured particle impact velocity, as seen in Fig. 6. It was stated that the wider velocity distribution led to poorer powder flowability and lower powder packing factor for these powder shapes.

Table 3 Hardness values for the 5083 substrate, NC powder and the NC deposit (Ref 40)

Material	5083 substrate	5083 powder	CSP 5083	
			Particle interior	High hardness zones
Hardness, GPa	1.13 ± 0.28	2.52 ± 0.36	2.77 ± 0.21	3.57 ± 0.32





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Fig. 6 Effect of CS particle shape on measured particle velocity distributions (Ref 66)

2.1.3. Surface oxide layer

Oxide films are almost always present on powder surfaces, and can strongly influence particle deformation in CS process [68, 69]. The critical velocity and DE are also influenced by the presence of a particle surface oxide film [70-73]. Pre-existing oxide surface films on powders can be disrupted and displaced from the interface by the outward flow of metal (metal jet). Nevertheless, cracked oxides generally persist to varying extents in CS deposits, with remnants often remaining at the interface after particle deposition [68, 74-76]. These oxides influence the metal jet and occurrence restoration phenomena (recovery & recrystallization) at the interfaces, and consequently the quality of the final deposit. Therefore, the effects of oxide films on particle deformation must be understood.

Li et al. [70] and Kang et al. [71] studied the role of oxide content on the critical velocity for CS deposition. Li et al. reported that when the oxide content of the Cu powder was increased from 0.02 wt. % to 0.38 wt. %, the critical velocity increased from 300 m/s to 610 m/s. In addition, Kang et al. [71] reported that the critical





velocity of Al powders increased by 125 m/s (from 742 m/s to 867 m/s) when the oxide content of the same powder size distribution was increased from 0.001 wt.% to 0.045 wt.%. The authors asserted that the critical velocity also depended on the thickness of the oxide layer, such that for thicker oxide layers, more energy is required to disrupt the layer and thus less plastic deformation is dissipated in the particle.

Yin et al. [77] studied the shape evolution of the oxide film on an Al 6061-T6 particle surface during the deposition process onto an Al 6061-T6 substrate. As shown in Fig. 7, the pre-existing oxide film crushes and fragments at the contact surface upon impact with the substrate. With a viscous-like metal jet extending outwardly, more cracked oxide is extruded from the interface (see Fig. 7(c-e)). Finally, the fragmented oxides are ejected to the surrounding space.

However, as shown in Fig. 7(f), the cracked oxides are not entirely removed from the interface after particle deposition. The same phenomenon has also been reported in experimental observations on CS deposition interfaces Al-Al [70, 75], Ti-Al [68], Cu-Cu [13, 76] and Cu-Q235 mild steel [72].



Fig. 7 Shape evolution of surface oxide film on an Al 6061-T6 particle during the CS deposition at an impact velocity of 700 m/s (Ref 77)



2.2. Geometric Effects

2.2.1. Spraying stand-off distance

Stand-off distance, defined as the distance between the nozzle exit and the substrate, affects particle velocity and temperature, and consequently affects particle deformation. Longer stand-off distances may decrease the DE and reduce property levels of CS deposits due to the profile of the velocity in the cross plane outside the nozzle. Optimal stand-off distance is generally distance from the nozzle at which particles reach maximum velocity. The effect of CS stand-off distance on particle deformation has been investigated experimentally [78], but more computationally [71, 79-86]. For example, Li et al. [78] showed experimentally that the DE decreased with increase of stand-off distance from 10 to 110 mm for Al deposits.

Numerical simulations have shown that optimal location of the nozzle in HPCS (to





reach maximum particle velocity), should vary with particle size and density [78, 79, 80, 83]. For example, based on Fig. 8, Li et al. [78] showed that for the same powder, the optimal stand-off distance during HPCS increases with increasing particle size. On the other hand, for different





powders of the same particle size, the optimal stand-off distance increases with increasing powder density.

In a numerical study, Patisson et al. [82] identified three distinct stand-off regions that could affect deposition performance: (1) the small stand-off region, where the presence of the bow shock, the shockwave at the substrate, adversely affects deposition (Fig. 9 (first row images)). This area is limited by the length of the supersonic portion of the jet, known as the nozzle supersonic potential core; (2) the medium stand-off region, where the bow shock disappears (Fig. 9 (second row left and middle images)) and if the gas velocity remains above the particle velocity (positive drag force), the DE continues to increase; and (3) the large stand-off region, where the gas velocity falls below the particle velocity (negative drag force), and particles begin to decelerate (Fig. 9 (second row right image)). For optimal deposition performance, the stand-off distance should be set within Region 2.



Fig. 9 Schlieren images of the bow shock at different standoff distances for 3.0 MPa and 20°C (Ref 82)





2.2.2. Spray angle

The effect of spraying angle on CS deposition has also been investigated in experiments [18] and numerical simulations [87-89]. Wipe tests have been widely used for investigating the effect of spray angle on particle deformation behavior [2, 3, 87]. In this test, the substrate moves fast enough from in front of spraying gun as though it ensures deposition of less than a monolayer of particles during a single pass through the spray jet. Schmidt et al. [18] and Binder et al. [87] showed the effect of spray angle on the microstructure of different CS deposits. They concluded that deviations from perpendicular impact conditions can significantly alter the particle deformation behavior, and usually leads to increased porosity and decreased tensile and bonding strength of the deposits (Fig. 10). However, they also showed that deviations from perpendicular impacts were $< 20^{\circ}$ (Fig. 10(c-d)), particles exhibit deformation behavior similar to normal incidence, with no significant changes in porosity level, which may be tolerable for most applications.



Fig. 10 Microstructures of CS Ti on AlMg3 substrates for different spray angles: (a, b) 90°, (c, d) 70°, (e, f) 45° (Ref 87)





Even when the spray angle is perpendicular, the particle reportedly can rebound or detach from the substrate after impact if the stand-off distance is not within region 2 (section 2.2.1) [77, 89]. Note that an additional temperature rise can occur at the interface in angular impacts due to frictional heating, which can facilitate shear instability. Also, the tangential component of the particle momentum in an angular impact creates a tensile force at the interface, which can suffice to detach the particle from the substrate (see Fig. 10(f)).

2.2.3. Position in particles jet

As particles exit the nozzle, they diverge in a conical spray as they travel towards the substrate. As a result, the in-flight velocity of the particles in the cone-shaped jet is non-uniform, particularly when particles undergo bow shock [82]. Guetta et al. [90] performed a thorough study on this effect, investigating the dependency of particle adhesion to the substrate as a function of particle position in the jet. Particles in and near the core of the jet have greater in-flight velocities, leading to greater deformation and penetration when they impact the substrate. The particle morphologies for different positions in the jet are shown in Fig. 11.

Based on these results, it can be suggested that using centralized nozzles can cause more particles in the jet to adhere or cohere during cold spraying and consequently, significantly increase DE.



Near the center

Intermediate position

Near the rim



2.2.4. Nozzle Geometry

As shown in Fig. 3, gas is supplied through a converging-diverging de Laval nozzle that accelerates the flow to supersonic. The nozzle geometry used in CS also influences the deformation behavior of particles by





controlling the supersonic flow in the jet. For any nozzle geometry, there is a maximum Mach number, and regardless of the operating pressure and temperature, the maximum Mach number is constrained by the nozzle geometry. To achieve a higher Mach number, the diverging section of the nozzle must be further expanded. Lee et al. [91] studied various nozzle geometries, under-expanded, correctly expanded (or shock-free), and over-expanded, *i.e.*, the air exit pressure is greater, equal, and lower than the ambient pressure, respectively. They concluded that overexpansion may allow surrounding (ambient) air to infiltrate the nozzle due to low static and stagnation pressure at the nozzle exit. This would adversely affect the flow by reducing flow velocities as momentum is dissipated through unwanted shocks. Across shocks, the Mach number and velocity decrease, although the density, entropy, static pressure, and temperature of the gas increase. Using numerical simulation Jahedi et al. [92] showed that an overexpanded nozzle is preferred over an underexpanded as it not only requires a lower nozzle chamber pressure but also produces a slightly greater impact velocity.

Overall, to have correctly expanded nozzle, where losses due to shock waves and shear interactions are minimized, operating pressure should be adjusted rather than operating temperature, as operating temperature has little effect on exit pressures. On the other hand, operating temperature is more effective than operating pressure at varying exit velocities.

2.3. Processing parameters

2.3.1. Carrier gas type, temperature and pressure

The particle velocity in CS is affected by the type of carrier gas used, the carrier gas temperature, and the carrier gas pressure. The effects of these variables have been investigated both experimentally [1, 3, 5], and numerically [18, 66, 87, 93-100].





In a recent study, Sue et al. [93] investigated the effect of carrier gas type on the particle velocity of different powder particles. Fig. 12 shows that use of He as a carrier gas can lead to velocity increases of 100 m/s or greater compared to other carrier gasses, *i.e.*, nitrogen (N_2) or compressed air. The results indicate that carrier gas species is critical to improving particle velocity and particle deformation, because high gas velocities typically yield greater DE values and better quality deposits. The reason for this behavior lies in the expression for local gas velocity shown as Eq. 2 [101]:

$$\upsilon = (\gamma RT/M_w)^{1/2}$$
 Eq. 2

where γ is the ratio of the constant-pressure and the constant-volume specific heat, which is typically set to 1.66 for monoatomic gases (He) and 1.4 for diatomic gases (N₂ and oxygen), respectively, *R* is the gas constant (8314 J/kmol·K), *T* is gas temperature, and M_w is the molecular weight of the gas. In this relation, the local gas velocity is positively correlated to $(\gamma/M_w)^{1/2}$, which is equal to 0.204 for argon, 0.220 for air, and 0.644 for He. Therefore, use of He as a carrier gas will yield the highest gas velocity and hence the highest particle velocity, followed by air and argon. However, He is also the most expensive and scarcest of these gasses, which has limited usage to specialized applications. At present, He recovery systems can be used during cold spraying. These systems help achieving 85-95% recovery, an achievement that could greatly expand the use of HPCS systems [28].







Fig. 12 ZK61 and copper particle velocities as a function of the carrier gas species (Ref 93)

In the HPCS process, the main gas is pre-heated before combining with the powder-gas mixture from the high-pressure powder feeder in a premixing chamber. The combination of gas and particles is generally injected axially through a de Laval type nozzle (*i.e.*, a converging-diverging nozzle), and the particles accelerate in the nozzle to impact on a flat substrate surface. This underscores the importance of carrier gas temperature on the particle impact temperature, particle deformation, and particle morphology after impact, and prior studies have addressed this issue [66, 71, 87, 91, 94]. To illustrate, Fig. 13 shows the effect of carrier gas temperature (or particle preheating temperature) on final microstructure of *Ti* deposits [87]. The particles undergo little deformation on impact in Fig. 16, T_{gas} =600°C, and deposited particles are roughly spherical. In contrast, increasing the process gas temperature to T_{gas} =1000°C causes the flattening ratio to increase markedly, and shear instabilities appear to be more pronounced (Fig. 13(b)). The flattening ratio (*R_f*) is defined as:

$$R_f = \frac{D}{d_p}$$
 Eq. 3

where D is the spreading diameter of the flattened particle, and dp is the original particle diameter.







Fig. 13 Morphologies of Ti particles sprayed under nitrogen gas pressure of 4 MPa (a) with Tgas = 600°C and (b) with Tgas = 1000°C (Ref 87)

Gas pressure plays an equally important role in particle impact velocity and particle deformation behavior. This assertion has been clearly shown by different authors [71, 97, 98, 102]. For example, King et al. [71] investigated the effect of gas pressure on Al powder particles deformation. Their results showed that the flattening ratio increased with gas pressure (Fig. 14).



Fig. 14 Ratio of bonds for Al powder with different oxygen contents at different gas pressures (Ref 71)





2.3.2. Substrate hardness, temperature and surface roughness

The characteristics of the substrate material also affect particle deformation during CS. These characteristics include material properties (hardness) and substrate surface preparation (surface roughness) as well as process parameters (substrate temperature). The relationship between these parameters and particle deformability of different materials has been reported extensively [77, 103-107]. For example, the effect of substrate hardness relative to particle hardness has been investigated experimentally by Bae et al. [103], Xiong et al. [104], and King et al. [106], and numerically by Yin et al. [107]. In general, the studies indicate that for a given impact particle velocity, and when the particle and the substrate are composed of different materials, reversing the particle and the substrate materials has a significant effect on crater diameter and on crater depth, as well as on the width of the interfacial jet (see Fig. 15).

For example, for the case of soft particle/soft substrate material pairs (Al/Al in Fig. 15(a)) extensive deformation occurs during impact due to the relatively low material strength of Al. In contrast, the hard particle/hard substrate case results in less extensive deformation (Ti/Ti in Fig. 15(b)). As shown in these two cases, the outer edge side of particle and substrate interface has a higher temperature than that of the center. Also, in both cases, the maximum temperature at the substrate side is greater than that of the particle side at the critical velocity. Furthermore, the Al/Al case shows a wider high-temperature region in the substrate than that of Ti/Ti, owing mainly to its relatively high ductility, high thermal-softening effect, and low strain hardening. Dissimilar combinations, e.g. soft particle/hard substrate (Al particle, mild steel substrate) and hard particle/soft substrate (Ti particle, Al substrate) are also shown in Fig. 15(c & d). The dissimilar combinations reveal distinct deformation behavior compared to the previously described cases. The initial kinetic energy of the particle is mostly dissipated into plastic deformation of the relatively soft counterpart. Accordingly, much higher temperature is achieved on the soft side. The simulation results by Bae et al. [103] demonstrate a flattened particle with a slightly deformed substrate for the soft particle/hard substrate case (Fig. 15(a)), and a





deeply penetrated substrate with a less deformed particle for the hard/soft case (Fig. 15(b)) at the critical velocity.



Fig. 15 Four different combinations of particle and substrate (Ref 102)

Substrate temperature can also strongly affect the deformation of impacted particles [108-111]. In one study, Legoux et al. [109] measured the DE as a function of surface temperature, and results are shown in Fig. 16. The DE for zinc decreases with substrate temperature while it increases for Al, and remains unchanged for tin. Thus, substrate temperature can be an important factor affecting particle deformation during CS processing, and it increases or decreases the DE depending on the materials deposited.



Fig. 16 DE as function of substrate temperature for different powder materials (Ref 108)





To date, the effect of substrate surface roughness has been a controversial subject in cold spray arena. Some data has shown increased roughness can enhance particle deformation and DE [111-113], while other data shows that substrate roughness reduces the DE and bond strength of the deposited material [114, 115]. Differences in surface profiles are represented in Fig. 17 (Hussain et al. [116]), which shows the line traces of both polished and ground surfaces. A representation of a 15 μ m Cu particle on the same scale is also shown. The polished surface is effectively flat, while the ground surface profile features 2-3 μ m perturbations, which can increase or decrease the extent of particle deformation experienced during impact.

Marrcco et al. [117] explored the effect of surface preparation techniques on the deformation of CS Ti particles deposited onto Ti6Al4V substrate. They proposed that grit blasting of the substrate causes work hardening, which subsequently limits substrate deformation during impact of the particles. They also argued that this restriction of substrate deformation results in less effective removal of surface oxide and thus leads to lower particle deformation. Wu et al. [118] studied CS Al-Si deposits onto both polished and grit-blasted mild steel. Micro-pores and defects were observed in the grit-blasted surface, while an "intimate" interface was observed following deposition onto a polished substrate. They asserted that micro-pores on the grit-blasted surface results in lower particle deformation. Similar behavior has also been reported by Yin et al. [111] for Ni deposits on polished and ground versus grit-blasted Al substrates.





Fig. 17 Surface profiles of polished, ground, and grit-blasted surfaces in comparison with a disk representing a section through a 15-Im-diameter spherical particle (Ref 115)

In contrast to these studies, Makinen et al. [119] report more extensive particle deformation for a Cu deposit cold sprayed onto a grit-blasted Cu surface compared to deposition onto an as-received surface. Also, Richer et al. [112] showed an increase in DE in spraying Al alloy particles onto a coarser grit-blasted surface compared to a finer grit-blasted surface. Overall, while these results are quite interesting and yet controversial, more detailed investigations would need to be conducted on the effects on substrate surface roughness on particle deformation before drawing concrete conclusions.

2.4. Key messages

Cold sprayed deposits are formed through impact and deformation of powder particles. Here, we reviewed the factors influencing particle deformation, and these were categorized into three groups - 1- powder characteristics (size, shape and state, surface oxide layer), 2- geometric effects (spraying stand-off distance, incidence angle, and nozzle geometry), and 3- process parameters (gas type, temperature and pressure, substrate hardness, temperature, and surface roughness).





These factors primarily determine particle temperature and particle velocity, and thereby affect particle deformation behavior. The key messages that can be drawn from published results are:

- Particle size influences particle deformation by causing changes to critical impact velocity. This occurs because of differences in heat capacity, thermal conductivity, cooling rate, and the degree of shear instability during impact for different particle sizes.
- 2- Particle state and shape strongly influence particle deformation behavior during CS due to variations in friction or drag coefficients.
- 3- Surface oxide films influences particle deformation by affecting particle velocity. However, oxide films on particle surfaces are disrupted during impact by the spreading flow of metal resulting in less oxides in the deposit than that of the feedstock powder.
- 4- Long stand-off distances decrease the DE and reduce property levels of CS deposits due to the profile of the velocity in the outside the nozzle.
- 5- Changes in spray angle lead to variations in frictional heating and temperature gradients at the substrate/coating interface and thus, affect particle deformation behavior.
- 6- Because particle velocity in the cone-shaped jet is non-uniform, particle position in the jet can affect particle velocity and associated deformation behavior. Centralized, correctly expanded nozzles can help to overcome this problem.
- 7- Particle velocity, and consequently the level of particle deformation, is strongly affected by carrier gas type, temperature, and pressure. Helium gives rise to the highest velocities for any powder.
- 8- Substrate hardness, temperature and surface roughness can change the maximum temperature increase at the interface and thus, affect particle deformation.





3. Microstructure of CS feedstock powder and deposition

Early investigations revealed that CS deposits retained feedstock microstructures [1, 2, 5, 18]. However, in recent years, microstructural characterizations have showed powder particle undergo SPD and features such as high dislocation density [18, 31, 32, 38-40, 74], ultra-fine grain (UFG) structures [41-46], phase transformation [54-56], and localized amorphous phase [57-60] have been observed. In this section, the evolution of these microstructural features in feedstock powders as well as in CS deposits are reviewed with respect to grain structure, dynamic restoration phenomena (recrystallization and recovery), precipitation, phase transformation, and amorphization. Understanding more clearly the relationships between the feedstock microstructure and CS deposit microstructures is critical to efforts to control product microstructure and optimize properties and performance of the deposit.

3.1. Microstructure of feedstock powder

The first step towards developing a clear understanding of CS microstructure is understanding the original feedstock powder microstructure. Here, we report salient features of the microstructure of gas atomized powder. Gas atomized powder is the most common type of powder used in the CS industry today, and vast majority of CS deposits are generated from this type of powder.

3.1.1. SEM observations

The appearance of a typical gas atomized powder is shown in Fig. 18 [39]. Fig. 18(a) shows that powder particle size can be slightly bimodal, consisting of a mixture of larger and smaller particles. Note that small particles below 5 μ m tend to agglomerate around larger ones during solidification, creating irregular shapes of the powder agglomerates. Fig. 18(b) shows a typical gas





atomized powder particle ~20 μ m in diameter. The ~1-4 μ m grain structure manifests as recessed grooves, along with smaller particles attached to the surface.



Fig. 18 SEM images of gas-atomized powder showing, (a) powder morphology, (b) surface structure (Ref 39)

Fig. 19 SEM micrographs of a typical gas-atomized showing cross section of (a) Type I particle showing internal grain structure and GB solute segregation and (b) Type II particle with GB precipitates (Ref 39)

Fig. 19(a) shows back-scattered electron (BSE) images of typical gas atomized powders, designated as Type I [39]. Type I particles exhibit a cellular structure often observed in metals solidified from the melt [31, 32, 38-42]. Cellular microstructure has been attributed to various mechanisms, including; (1) high cooling rate and rapid solidification [120], (2) pre-solidified microdroplets and dendrite fragments [121], and (3) thermal equilibration and partial re-melting of solid particles [122]. Note that the BSE image in Fig. 19(a) shows strong z-contrast (atomic





number contrast), associated with composition variation at the grain boundaries (GBs) resulting from solute segregation during solidification. Solute segregation has been widely observed in gas atomized powders [31, 39, 123-127]. However, Type II particles are also found in some of the gas atomized powders within the same batch, as shown in Fig. 19(b) [39]. The difference between the two types lies in the internal grain and GB structure. In contrast, Type II particles exhibit coarser grains with more extensive GB precipitation compared to Type I. Type II particles may experience slower cooling rates during atomization, affording sufficient time for GB solute to precipitate. Researchers [124, 125, 127] have argued that the Type I microstructure is observed mostly in smaller particles, while Type II is typical of larger particles.

Note that not all powder batches feature both particle types. If the feedstock powder is a precipitate hardened alloy, both particle types are prevalent throughout the powder batch in ratios of ~ 3:1, Type I to Type II [39]. On the other hand, if the spraying powder is not a precipitate hardened alloy, then the Type II is normally absent. Differences in powder microstructures may or may not be significant for achieving a given coating property, but it is important to consider that microstructural variations in the powder will also be present in the CS deposit.

3.1.2. In-depth characterization

EBSD and TEM have been used to understand the internal grain structure of powder particles [39, 46, 128]. A typical interior grain structure, in this case, a gas atomized Al7075 particle, is shown in Fig. 20(a) [39]. The interior grain structure resembles the surface grain structure shown previously (Fig. 18(b)). The image in Fig. 20(a) also shows that the grain structure is not uniform, with asymmetrical grain shapes and micron-size grains.







Fig. 20 (a) EBSD and (b) TEM images showing the presence of dislocation structures and LAGBs (white lines in (a) in a gas atomized 7075 Al powder. The inserted SADP is from the same region as in (b) (Ref 39)

Using EBSD (Fig. 20(a)), it was confirmed that the grains in feedstock powders are also not fully relaxed, and internal stress is present in the powders [39]. The particles show regions with a high and low densities of low-angle grain boundaries (white lines in Fig. 20(a)), which presents evidence of dislocation substructure existing in the as-received powder. The presence of dislocation structures was also observed in TEM images (Fig. 20(b)). This image shows that the feedstock powder particles contain dislocation substructure. Many grains have a complex diffraction contrast and the inserted selected area diffraction pattern (SADP), collected from the entire area in Fig. 20(b), shows several spots which are elongated.

3.2. Thermo-mechanical processing mechanisms in CS deposits

Because of the high-velocity impact during CS, particles experience large strains (up to 10) and strain rates up to 10^{9} /s, particularly in interfacial regions [46, 51, 128-131], leading to work hardening. Because of the severe





SPD and the associated temperature rise, restoration phenomena occur, specifically dynamic recovery (DRV) and dynamic recrystallization (DRX). These phenomena play important roles in the formation of new microstructural features in the deposit.

The schematic in Fig. 21 depicts the formation of UFGs based on the adiabatic shear instability (ASI) and DRX [43]. When a particle strikes the substrate (a), the high impact pressure generated at the impact region and the associated intense shear stress deform the splat. As the contact area between particle and substrate increases, ASI occurs, grains undergo SPD, and a material jet occurs by viscous flow. Next, grains in the contact region form dislocation cells (b), and finally subgrains develop and elongate (c). If the strain and the temperature exceed critical values, the subgrains rotate and recrystallize due to SPD and the associated thermal softening, triggering viscous flow (d). Similar grain-refined regions reportedly distribute along particle boundaries in CS deposits of Al, Cu, Ni, and Zn [57, 74].



Fig. 21 Schematic evolution of grain refinement by dynamic restoration phenomena in CS deposits (Ref 43)





3.3. Microstructure characteristics of CS deposits

3.3.1. General features

LM and SEM have been used to investigate the morphology of deformed particles in as-deposited CS materials [58, 75, 132-137]. In these characterizations, the CS deposits are dense, and the porosity is typically negligible (< 1%), as shown in Fig. 22. The image also reveals the presence of prior particle boundaries (PPB's), a common feature of powder metallurgical (P/M) structures [138, 139]. Each particle experiences sufficient plastic deformation to conform to the underlying deposited layer, resulting in a dense deposit.



Fig. 22 Etched cross sections of cold-sprayed Cu coatings (Ref 134)

CS deposits often exhibit a heterogeneous distribution of deformation within deposited particles. One of the earliest microstructural characterization studies of CS Al deposits by Morgan et al. [140] showed that a microstructure comprised of highly deformed and more lightly deformed regions (Fig. 23), that corresponded to peripheral and interior regions of particles, respectively. As shown in Fig. 23, the level of deformation varies with radial position within the PPB's, with




peripheral regions showing extensive deformation, and interior regions showing limited deformation. This heterogeneous deformation leads to bimodal grain structures in CS deposits. Similar microstructural features have also been reported for other CS deposits [31, 39-42, 46, 128, 140].



Fig. 23 Deformation of grain structure within the particle at the leading edge (Ref 139)

The general features of the microstructure at the scale of the splat/particle, described above, raise questions about what occurs at a finer scale during the build-up process. The corresponding phenomena govern the final (mechanical and physical) properties of the coating, described in the following sections.

3.3.2. Fine-scale microstructure

Details of the grain structures within the two regions described in the previous section have been revealed using EBSD [31, 46, 128, 141-144]. For example, Zou et al. [46] reported inhomogeneous deformation during cold spraying of Ni and Cu, observing a mixture of UFGs and elongated grains of different sizes in the microstructure (Fig. 24(a)). Rokni et al. reported similar





features in different CS Al alloys [31, 32, 38, 41, 144]. As shown in Fig. 24(b), particle-particle interactions caused the formation of UFGs (dashed circle) in peripheral regions of highly deformed particles in CS 6061 Al deposits [144]. A third region, where the degree of plastic deformation is not as great as in UFG regions, was also observed in the CS deposit, and these are aptly termed "pancake structure" (black arrows in Fig. 24). The flattened, elongated grains reflect extensive shear deformation but do not qualify as fine-grained [31, 32, 38-41, 144].



Fig. 24 EBSD characterization of the cross section of (a) Ni (Ref 46) and (b) 6061 Al deposits (Ref 143)

The evolution of microstructure within CS deposits and particularly the pancake grain structures at some particle interfaces, has also been revealed by TEM (Fig. 25). The mechanism by which such pancake grain





structures form has been attributed to ASI on particle surfaces during high velocity impact [2, 3, 5, 18, 32, 46, 61, 74, 128]. Crystallographic slip on a dominant slip system propagates through multiple- or cross slip events, producing the distinctive grain shape [145]. Indeed, nearly identical grain shapes are also observed in adiabatic shear bands produced by dynamic loading [146-150]. The white lines in Fig. 25(a&b) outline the pancake grains and highlight that they partition into the "ladder-like" structure with low-angle grain boundaries (LAGBs).



Fig. 25 TEM micrographs of CS (a) 7075 and (b) 6061 showing pancaked structures with interconnecting LAGBs, dashed lines show ladder-like structure (Ref 143)

Lee et al. [158] illustrated the typical microstructure between two CS deposited particles in Fig. 26, showing the features common in different CS deposits. The illustration assumes that particles experience sufficient plastic deformation to form equiaxed UFGs at the particle-particle interface (region (1)). Pancake grain structures with LAGBs dividing them into ladder-like structures appear next to the UFGs (region (2)). Adjacent to the pancake grains, moderately deformed particle interiors with dislocation walls or/and cells formed in these regions (region (3)).







Fig. 26 Schematic illustration of microstructure at inter-particle boundaries in CS deposits (Ref 157)

Polished sections through individual CS particles provided further insights into the distribution of plastic flow and the evolution of grain structure. King et al. [159] used focused ion beam (FIB) to section through a single CS Cu particle, shown in Fig. 27. There was a general tendency for increasing deformation with increasing subsurface depth (below the top surface of the particle), with the maximum deformation occurring near the particle-substrate interface.



Fig. 27 FIB image of the cold-deposited particle cross section, and areas chosen for analysis of grain deformation (Ref 158)





Various other features associated with heavy deformation have been revealed in CS metallic microstructures via TEM, including dislocation cell structures and dislocation loops [41, 42]. The significant dislocation density in CS deposits is consistent with SPD at high strain rates during processing. Compared with a typical dislocation density of ~ 10^{12} m⁻² in annealed alloys, the density can be increased up to 10^{16} m⁻² for cBN/NiCrAl nanocomposites [160].

3.3.3. Formation mechanisms of the UFG structures

The appearance of the typical UFG structures resulting from CS is shown Fig. 28, which shows an EBSD orientation map of a CS 7075 Al deposit showing the complexity of grain structure. Note that some particles experience sufficient deformation and heat during impact to completely recrystallize, producing UFG structures (dashed circles). There are few LAGBs in the UFG regions compared to surrounding areas. Enlargements of these regions are shown below in Fig. 28. The peculiar grain structure has been cited as evidence for the occurrence of continuous DRX (CDRX) and conversion of the LAGBs to HAGBs. Indeed, dynamic recrystallization via CDRX has been reported during deposit of other metals and alloys [39, 46, 144].

Geometric DRX (GDRX) has also been proposed as a mechanism for formation of UFG structures in high stacking fault energy (SFE) alloys, particularly Al alloys, deformed under large strains at high temperatures (> 200°C) [39, 144, 145, 161-165]. Pinching off and annihilation of HAGBs occurs as the original grains thin to about a few times the subgrain diameter [163, 164]. The GB annihilation is due to the presence of warm or hot high-strain deformation that can result in serrated boundaries, presumably in association with DRV [163]. High localized deformation at PPBs, produced by CS processing, may thin the grains (observed in Fig. 24) to the dimensions of the subgrain diameters, producing UFG structures. As such, the occurrence of GDRX in regions





with pancake grain structures (black rectangle in Fig. 28) can transform these regions into UFG structures (white rectangle), as described elsewhere [39, 144].

A mechanism known as rotational DRX has also been proposed [46] as a possible cause for DRX adjacent to interface regions in CS deposits. In the case of rotational DRX, strain-free grains are created in another way. The sequence of rotational DRX is shown in Fig. 29. This mechanism, which was originally reported by Zou et al. [46], has been explained by Kim et al [158].



Fig. 28 EBSD pattern from the 7075 Al CS deposit indicating the LAGBs distribution. The magnified images show the yellow dashed circles pointing to less LAGBs in these areas (Ref 39)



Fig. 29 Sequence of rotational DRX: (a) formation of elongated subgrains due to an accumulation of dislocations, (b) breakup of elongated subgrains, and (c) the rotation of the broken subgrains (Ref 46)





Static recovery (SRV) and static recrystallization (SRX) can be assisted by the residual heat of plastic deformation occurring in subsequently impacted particles [44, 158, 166]. These restoration processes are more likely to occur when heat energy is stored in the microstructure of the deposited material due to subsequent particle impact. When SRV and SRX occur, the microstructure alteration occurs in the highly strained areas. As schematically shown in Fig. 30, both static recovered and static recrystallized microstructures appear in a CS Al deposit. However, because SRV and SRX are competitive phenomena, most stored strain energy is consumed first by SRV, and if the strain energy is sufficient to activate recrystallization, SRX will consume the remainder of the stored energy, generating recrystallized grains [158, 166]. Note that in high SFE materials such as Al, the activation energy required for SRV is lower than for SRX and the dislocation glide and climb is active. Therefore, most of the stored strain energy was relieved through SRV [145, 167-170].



Fig. 30 Sequence of microstructural change via SRV and SRX during the kinetic spray: (a) microstructural state after the deposition stage, (b) additional heating by subsequent particle impact, and (c) microstructural change via SRV and SRX (Ref 165)

The extent and distribution of UFGs, or simply the degree of recrystallization, depends on the material properties of the feedstock powder. Differences in SFEs, activation energies for recrystallization, and thermal conductivities can affect the extent of recrystallization. For example, Zou et al. [46, 128] reported less uniform recrystallization for Ni compared to Cu, and Bae et al. [44] observed that in the case of poor thermal conductors such as Ti, SRV and SRX can be much more extensive due to local retention of transient thermal energy. Fig. 31 shows grain refinement





has transformed about half of the splat (by volume) into UFGs through recrystallization, primarily SRX.

In cases where the starting powder is nanostructured, the NC structure is often retained after CS, as observed in a cold-deposited Al-Mg alloy [40,171]. There have been reports of dynamic amorphization in Al-Al, Ni-Al [58], Al-Cu [172], Cu-Ni [173], Al-Mg [174, 175] and Fe-Al [176, 177] systems, over thicknesses of up to tens of nanometers. In the case of metallic glasses, CS can result in partial devitrification of the initially amorphous phase [177]. A wide range of examples of amorphization and dynamic recrystallization in CS deposits have been reported [46, 158, 172, 177, 178].



Fig. 31 TEM image of a thin foil of a Ti cold-deposited splat onto a Ti-6Al-4V substrate (Ref 158)

3.4. Effect of post-CS heat treatment

Although the CS process involves SPD of the powder particles, the plastic strain experienced by particles is non-uniform and thus the microstructure of the deposited material is also non-uniform. However, heat treatment (HT) can change and homogenize the microstructure of CS deposits and generate different microstructural features throughout the specimen. As a result, HTs have important implications for microstructural homogeneity and optimization of mechanical properties in CS deposits.



Fig. 32 shows typical cross-sectional views of stainless steel (SS) 316L as-deposited and heat-treated coatings produced using N_2 as propellant gas [141]. In the as-deposited condition (Fig. 32(a)), particles have high aspect ratio and the deposit contains some porosity and triple junction voids due presumably to incomplete bonding of particle-particle interfaces. A decrease in porosity was obtained in the deposit after annealing at 400°C, as shown in Fig. 32(b). Also, the splat shaped particles are now replaced with more equiaxed grains. There is a marked decline in porosity level after annealing at 1100°C (Fig. 32(c)). The decrease in porosity with increasing annealing temperature in the N₂deposited coatings was explained using sintering models. The fine porosity observed in the CS deposit after annealing was also attributed to incomplete "sintering" at inter-

particle interfaces have not been metallurgically bonded. Fig. 33 shows a montage of TEM images from a CS 7075 Al deposit after annealing [31]. The annealing process provided some additional thermal driving force for SRV and SRX, releasing much of the stored





Fig. 32 SEM images of cold-sprayed SS 316L coating deposited by N2 in the (a) as-coated condition, (b) after HT at 400 C, and (c) after HT at 1100 C (Ref 140)

energy. The grains are equiaxed with well-defined, straight GBs, unlike the inhomogeneous microstructure observed in the pre-annealed deposit (Fig. 24). Furthermore, many of the grains are free of dislocations (limited diffraction contrast) and residual stress has been relieved through the





annealing process. In these cases, the grains appear dislocation-free because the slip distance across the UFG grains is small enough that dislocations can rapidly glide across the grain during annealing and be absorbed into the opposite GB without major interactions with each other [40]. This mechanism has also been reported to play an important role in other SPD materials, such as cryomilled UFG Al extrusions [179-181].



Fig. 33 Montaged TEM image obtained from the CS 7075 deposit after annealing at 450 C for 45 min (Ref 31) As described before, the distinctive microstructure of CS deposits contains three distinct regionsparticle interiors, particle interfaces with UFG structure, and particle interiors with pancake grain structure. The microstructure in these three regions after annealing can be compared against the pre-annealed deposit. It was also observed that interior regions of 7075 Al CS particles did not significantly change after annealing (Fig. 34(a)) [31]. This retention of grain structure has been attributed to the presence of fine age-hardening precipitates within grains and their effect in restricting GB migration [182-184]. However, as shown in Fig. 34(b), peripheral regions of particles with pancake grain structures exhibit recrystallization during annealing, and develop a UFG structure. Peripheral regions of particles that were originally characterized by UFG structures prior to annealing, also are relatively unaffected by annealing (Fig. 34(c)). Although subtle, these regions showed a slight reduction in overall dislocation density, increased GB definition, and





evidence of modest recrystallization.

In all reports of CS deposits and associated microstructures, negligible post-annealing grain growth has been consistently reported. Resistance to grain growth in CS deposits has been attributed to the extensive solute segregation at GBs in the deposits [32, 39, 125, 127, 182-184]. The segregation can lead to nucleation of GB precipitates during annealing and restrict GB

migration. Due to the nature of the gas atomization process, solute segregation to GBs in gas-atomized powders is almost unavoidable, as discussed in section 3.1.1. However, more extensive GB segregation has been reported in as-deposited materials given the small grain size, short diffusion distances, and transient heat of deformation during CS.

In some cases, blending different powders, such as SS 316 with Co–Cr alloy L605 powder [142] before cold spraying, and then performing post-CS heat treatment, produces composite deposits with superior corrosion and mechanical properties compared to the unblended powder (316L alone). The low deposition temperature, high deposition rate and relatively low cost make CS an efficient process for fabrication of composite deposits [5, 7, 19, 22, 23, 185, 186]. CS of Ti/Al [79, 187], Zn/Al [188], Fe/Al [189], and Ni/Al [190] powder blends show that dense composites (Fig. 35)



Fig. 34 TEM micrographs obtained from different regions in the CS 7075 deposit after annealing to 450 C, including (a) particle interiors, (b) particle interfaces with the pancake structures, and (c) particle interfaces with UFG structure (Ref 31)





can be produced using blended powder, and subsequent annealing can lead to intermetallic (IM) phase formation.



Fig. 35 Cross-sectional microstructures of Fe/Al composite coating annealed at (a) 450 C, and (b) 600 C for 4 h showing the formation of Fe/Al IM compounds (Ref 187)

Al/Mg CS deposits are also commonly heat-treated after deposit to generate the IM phases Mg₁₇Al₁₂ (β) and Al₃Mg₂ (γ), increasing hardness and corrosion resistance [191-194]. Bu et al. [191] reported four distinctive zones in CS Al/Mg deposits after 4 h heat treatment at 400 °C, as shown in Fig. 36(a). The top zone is the unreacted as-sprayed Al deposit, while the second and third zones represent IM layers formed at the coating/substrate interface. The fourth zone is the AZ91D magnesium substrate shown at bottom. These observations were accompanied by EDS analysis (Fig. 36(b)) that showed that the IM compound near the substrate was Mg-rich phase (β phase), while the IM layer near the un-reacted Al deposit was the Al-rich γ phase.



Fig. 36 (a) SEM image of CS Al onto AZ91D-T4 after HT under vacuum for 4 h at 400 C revealing different IM phases, and (b) quantitative EDS line scan as indicated by arrow in Fig. 36(a) (Ref 189)

Like any powder metallurgy product, the sintering/annealing environment is critical to the microstructure and mechanical properties of CS deposits. Vacuum or oxygen-free atmosphere (*i.e.*, argon (Ar) or nitrogen (N₂)) for HT are generally performed to prevent the formation of oxide layers on the deposit surface and to avoid the formation of sintering necks. Lee et al. [195] investigated the effect of different HT gas environments on CS Ti deposits. Fig. 37 shows that vacuum annealing yielded superior densification (3.8% porosity, 156.7 HV) relative to Ar annealing (5.3%, 144.5 HV) and 5%H₂+Ar gas annealing (5.5%, 153.1 HV). The results also revealed that vacuum environment during post-CS HT can reduce oxide content (purification) in the Ti deposits.



Low annealing temperatures can avoid distortion and residual stress development and most importantly, preserve the benefits of using a low-temperature deposition technique. These benefits include overall compressive residual stress, fine microstructure, and superior hardness. Moreover, even if beneficial, a post-deposition anneal can be difficult or unacceptable from a practical perspective because of the component size or base material or even specific production process.

3.5. Key messages

As-atomized powders generally contain two types of particles, those with cellular microstructures, and those with extensive GB precipitation, which tend also to be larger. If the source powder is not a precipitation hardened alloy, then the second type of particle – those with GB precipitation -are normally absent. Particles experience high levels SPD during the CS process, resulting in high-density deposits with low porosity, though un-joined interfaces are occasionally induced at particle-particle boundaries. CS deposits often exhibit non-homogeneous distribution of deformation within deposited particles, and the microstructure generally consists of three distinct regions: 1- particle interior regions; 2- particle-particle interface regions containing pancaked grains, and 3- particle-particle interface regions containing an UFG structures generated through recrystallization. The primary recrystallization mechanisms are CDRX, GDRX, subgrain rotational, and SRX. The extent and distribution of UFGs, or simply the degree of recrystallization, depends on the material properties of the feedstock powder, particularly SFE and





thermal conductivity. Phase transformations and chemical reaction rarely occur during CS because the particles are not heated directly by a thermal source. However, post-CS annealing can form not only IM compounds, but also can create more homogenous microstructures with superior mechanical properties. Composite deposits can also be produced via CS and limits to the versatility of CS have yet to be encountered.

4. Mechanical properties of CS deposits

4.1. Overview of mechanical properties related to deposit quality

The mechanical properties and structural integrity of CS deposits are generally one of the main issues for developing CS processes for different industries, like defense, aerospace, and automotive, in which CS is utilized as a repair/refurbishment process [161, 196, 197]. These industries often have demanding requirements for the mechanical properties of deposits. As a result, HPCS is finding increasing use for many structural applications, due to the higher strengths, densities, elastic moduli, and ductility exhibited by CS deposits compared to LPCS. Additionally, CS offers unique advantages over other thermal spray processes, including lower temperatures and the unique solid-state microstructural growth mechanism, which can result in distinct microstructures. The plastic deformation that occurs during impact, and the associated heating and cold working of additional deposited layers, as well as post-CS heat treatments, can further alter the microstructure, opening new opportunities to control and enhance the resultant mechanical properties of CS deposits. In this section, we review fine- and bulk-scale mechanical properties of CS deposits and the effects of post-CS annealing on mechanical properties. Our goal is to provide a useful summary for CS experts as well to guide students and industrial end users to achieve a rapid understanding of the process characteristics and potential for specific industrial applications.





4.2. Local mechanical properties

Deformation inhomogeneity in CS deposits creates different regions in the microstructure (Section 3.3.1). As a result, these regions affect local mechanical properties of CS deposits. As discussed previously, particle-particle interfaces are often 1-4 μ m wide. Therefore, nanoindentation is well suited to characterize local variations in mechanical properties in CS deposits, owing to the fine resolution in load (~ 1 mN) and displacement (~ 1 nm) [19, 32, 40, 46, 126, 136, 171, 198-201].

4.2.1. Nanoindentation and modulus

Nanoindentation has widely been used to measure the elastic modulus of CS deposits [200], as well as the hardness of as-received metal powders [104, 110] and CS deposited pure metals and engineering alloys [32, 40, 45, 46, 171, 199, 202, 203], and composite materials [204-206]. Furthermore, nanoindentation studies have reported variations in local mechanical properties of CS deposits [32, 40, 46, 128, 198].

Because of grain size differences in the distinct microstructural regions, *i.e.*, particle interiors, PPBs with pancake structure, and PPBs with UFG structures, almost all studies have reported a nonuniform hardness distribution within CS deposits. Soer et al. [207] reported that, generally, the hardness increased towards particle interfaces and measured 0.7 and 0.8–1.2 GPa higher hardness for particle interfaces than the particle interiors in Fe–14%Si and Mo, respectively. Zou et al. [128] and Rokni et al. [32] used EBSD along with nanoindentation to measure the local variations of mechanical properties, as shown in Fig. 38. They both reported the same trend, higher hardness at particle-particle interface regions, in Cu [128], Al alloy [40, 46, 128, 198] deposits. The mechanisms considered included 1- GB strengthening (lower grain size), 2- precipitate strengthening, and 3- strain hardening, and were treated [128] using linear superposition. The





increased hardness near PPB's was attributed to the increased density of GBs and dislocations at these regions. GB strengthening and strain hardening induced by CS were the primary mechanisms responsible for the increased hardness at PPB's.



Fig. 38 EBSD characterization and nanohardness distribution maps of: (a, b) Ni (Ref 127), and (c, d) 7075 Al deposits (Ref 32)

A similar trend was reported for NC CS materials. Ajdelsztajn et al. [201] measured a hardness of ~ 4 GPa for cryomilled 5083 Al coatings deposited by cold spray. In another study, Rokni et al. [40] observed hardness variations between the 5083 substrate (as-cast condition), the NC cryomilled powder, and different regions in cryomilled NC deposits. As shown in Table 4, the average hardness of the NC powder was greater than that of the as-cast substrate. Also, the CS NC deposit showed higher hardness than the feedstock NC powder, with the greatest hardness at PPB's. The results indicate that milling/mechanical alloying processes combined with the CS technology may be a viable means of producing both NC deposits and bulk nanostructured materials for engineering applications.

 Table 4 Effect of thickness on bond strengths of pure Al/2024Al deposits (Ref 227)

Deposit thickness,	Average bond strength		Type of
mm	MPa	SD	failure
0.152	43.11	±3.86	Adhesion
0.508	39.56	±5.01	Adhesion





4.3. Bulk mechanical properties

Bulk mechanical properties of CS deposits affect the performance of repaired/refurbished parts and components which are re-used in their applications. Deposition parameters, including gas temperature and pressure, particle velocity, and substrate material, alter the CS microstructure and cause variations in bulk-scale mechanical properties of CS deposits. Therefore, the relationship between CS parameters and final properties provides important insights.

4.3.1. Microhardness

Uniformity of mechanical properties across CS deposits affects structural integrity in CS repair parts and components. Microhardness measurements provide a useful means to assess the uniformity of basic strength properties. Substrate materials and CS deposits generally show different microhardness values [32, 129, 208, 209]. When the substrate and CS deposit are the same material, the difference in hardness between substrate and CS deposit stems from three major factors: 1- difference in tempers (for age-hardened alloys) [32], 2- work hardening because of SPD, and 3- presence of porosity in the deposit [32, 209]. In general, however, the hardness of CS material is usually comparable with that of the same bulk material in a similar material condition. In the case of dissimilar substrate/deposit materials, the hardness of the deposit and the substrate often differences.

Ghelichi et al. [113] investigated dissimilar and similar combinations by depositing Al7075 and pure Al on pure Al substrates. Microhardness measurements across the deposits and substrates showed that the deposits of Al7075 had twice the hardness of the pure Al (see Fig. 39(a)). On the other hand, the substrate surface hardness of specimens coated with pure Al showed a lower microhardness compared to the former series because of greater deformation in the coating/substrate region.





Fig. 39 Microhardness distribution from the top of the (a) CS 7075 and CS Al on Al substrates (Ref 112), and (b) CS 7075 Al on 7075-T6 substrate (Ref 32). The dashed red lines represent the location of the deposit/substrate interface

Rokni et al. [32] investigated the combination of CS 7075/7075-T6 substrate (Fig. 39(b)) and reported that the CS deposit hardness $(115 \pm 9 \text{ Hv})$ was much lower than the substrate hardness $(171 \pm 5 \text{ Hv})$. This observation was attributed to the temper difference between the substrate (T6) and both 7075 gas atomized feedstock powder and the CS deposit (solution treated). They pointed out that, by comparison, the CS deposit microhardness was slightly greater than the reported hardness for solution treated 7075 Al (W temper, 105 Hv) due to the SPD during CS. The hardness distribution across the deposit was divided into three distinct regions (Fig. 39(b)). Region I, near the surface of the CS deposit, exhibited the lowest microhardness values because of internal porosity and poor bonding [39, 171]. This observation also indicated less extensive plastic deformation (during deposition) in the final CS layers, which was associated with the absence of the so-called peening effect [209-213] in this region. The microhardness was slightly greater in Region II, although significant strain hardening was expected in this area. Region III on the other hand, showed a significant





increase in hardness in the area close to the coating/substrate interface, approaching the substrate hardness. This increased hardness derived from a smaller grain size, which resulted from the shot peening effect and larger plastic strains. Fig. 40 shows EBSD maps obtained from Region I, II, and III and the corresponding grain size measurements. There was a steady reduction in grain size from the surface of the deposit to the interface.



Fig. 40 EBSD maps from (a) top (region I), (b) middle portion (region II), and (c) bottom portion (region III) of CS 7075 Al deposit. Upper right corner boxes show the average grain size in these regions (Ref 32)

Furthermore, virtually all CS parameters affecting particle deformation behavior (reviewed in Chapter 2) can change deposit hardness. However, there are some discrepancies in the literature regarding the relationship between CS parameters and microhardness. For example, Goldbaum et al. [48] showed that microhardness values for spherical and non-spherical deposits were similar for all spray conditions and not significantly greater or less than the hardness of bulk Ti, as shown in Fig. 41(a). On the other hand, Wong et al. [66] reported that powder state and shape affected the resultant hardness of CS deposits (Fig. 41(b)), and attributed this to the level of porosity in the corresponding deposits. In general, sponge and irregular powders show the greatest hardness in deposited materials for different V_p/V_{cr} ratios, as shown in Fig. 41(b). Note that the same





trend has been reported for deposits built from cryomilled NC- vs. gas atomized powders [126]. A significantly higher hardness has been achieved in deposits with as-cryomilled powder, since NC grains are produced by mechanical milling under liquid nitrogen and the grain size range is markedly smaller than as-atomized powder.



Fig. 41 Microhardness difference for (a) various deposition velocity and powder states for Ti deposits (Ref 48), and (b) different powder states (Ref 66)

Regarding the effect of particle size on the hardness of deposited materials, Zahiri et al. [214] showed that deposit microhardness decreases with decreasing average particle size (22 to 16 μ m) under similar spray conditions for Ti deposits. However, Marrocco et al. [117] and Cinca et al. [215] showed the opposite trend - deposit microhardness decreased with increasing average particle size (28 to 47 μ m). Thus, it appears that the effect of average particle size on deposit properties may be dependent on the range of the size distribution.





4.3.2. Bond strength

Adhesive strength or bond strength of CS deposits affects industrial applications. In general, average adhesive/cohesive strengths are determined by gluing the sprayed area to a respective counter-body of the same size, then pulling the assembly in tension to failure (ASTM C633). Using this method with certain material combinations has shown that deposits can fail cohesively [133] or adhesively [21]. There are cases where the test results are limited by the strength of the adhesive. To overcome this case, others [216] have attempted alternative testing techniques which have indicated that CS deposits can display high adhesive strengths (up to 250 MPa for Al alloys [17]).

Bond strength has been investigated by assessing the effects of parameters such as particle velocity [216, 217], carrier gas temperature & pressure [48, 87, 135, 218, 219], stand-off distance [78, 220], surface roughness [11, 117, 221], and spray angle [18, 87] on the quality of CS deposits. The effects on some of these parameters are reviewed below.

Greater bond strength values are measured in the CS deposits sprayed with higher gas temperature and pressure [48, 69, 87, 117, 135, 216, 219], as shown in Fig. 42. Greater bond strengths have been attributed to two factors. First is the higher kinetic energy of the particles at higher gas pressures, which increases particle deformation [87]. Second is the greater ductility of the spray material at higher temperature [5, 18, 31], which also increases particle deformation. Note that the effect of gas temperature reportedly is more effective than gas pressure on bond strength, especially with N₂ gas [219]. By increasing the gas temperature, both particle and substrate are softened by heating. Particles that deeply embed into the substrate increase the material mixing and interlocking at the interface, giving rise to superior interface bond strength.





Fig. 42 Effects of carrier gas temperature and pressure on the bond strength of Ti deposits (Ref 48)

Carrier gas type can also affect the final bond strength of CS deposits. According to Stoltenhoff et al. [135], the adhesion strength between Cu and SS substrate increased when He was used as a carrier gas instead of N₂ (marked with blue stars in Fig. 43(a)). Using He as a process gas, particles work-hardened more extensively because of greater particle velocities, producing greater bond strength between deposit and substrate [222]. Substrate material also affects bond strength [135, 216]. As shown in Fig. 43(b), the bond strength of Cu deposits on Al5052 and Al6063 substrates is greater than on a Cu substrate. These results confirm that a greater particle velocity results in more extensive plastic deformation for the Al substrates compared with the Cu substrates in order to form an effective bond.

Fig. 43

There is general agreement that grit blasting of substrates can enhance mechanical interlocking of CS deposits to substrates by increasing surface roughness [223-227]. Sharma et al. [227] investigated the effect of different surface preparation methods on final bond strength of pure Al deposits on 2024 Al substrates. As shown in Fig. 44(a), the greatest mean adhesive strengths were achieved by SiC grit blasting at 45° using N₂ carrier gas and the glass bead blast at 90°C using He gas. However, some discrepancies have been reported this regard. For example, Marrocco et al. [117] reported lower bond strength between Ti deposits and substrates (Fig.





44(b)) after grit blasting the substrate, while the polished and ground surfaces resulted in greater bond strengths.



Fig. 44 Effect of surface preparation on bond strength of (a) CP-AI/AA2024-T3 (Ref 225) and (b) Ti/Ti (Ref 116) deposits Spray angle can also affect the bond strength of CS deposits. The correlation shown in Fig. 45 (Binder et al. [87]) confirms that the number of well-bonded PPB's decreases with decreasing impact angles. The bond strength decreases from ~290 MPa for perpendicular impacts, to ~90 MPa for a spray angle of 45°, demonstrating that impact angles often have a stronger influence on deposit bond strength than other CS process parameters.

CS deposit thickness reportedly influences bond strength. Fukanuma et al. [216] and Sharma et al. [227] demonstrated that the thicker the deposit, the greater the risk of de-bonding between deposit and substrate because of the accumulated internal stress, as presented in Table 5. Sharma et al. [227] explained that thinner coatings have less residual stresses, contributing to the slightly greater adhesive strength.





Fig. 45 Shear strength of CS Ti deposits on low carbon steel substrate for different spray angles and carrier gas temperatures performed at a carrier gas pressure of 4 MPa (Ref 87)

	Material	Gas	Gas temperature, °C	Gas pressure, MPa	Electrical conductivity, % IACS
(Ref 128)	Pure copper	Air	450	2	~ 31
(Ref 244)	Pure copper	Air	450	2.2	35 ± 3
(Ref 245)	Pure copper	Nitrogen	350	3	50
(Ref 18)	Pure copper	Nitrogen	300	3	60
(Ref 134)	Pure copper	Nitrogen	305	3	62
(Ref 18)	Pure copper	Nitrogen	600	3	80
(Ref 246)	Pure copper	Nitrogen	600	2.7	
(Ref 245)	Pure copper	Nitrogen	800	3	80
(Ref 18)	Pure copper	Nitrogen	800	3	80-90
(Ref 18)	Pure copper	Nitrogen	800	4	90-95
(Ref 241)	Pure copper	Nitrogen	900	3	90.7 ± 5
(Ref 232)	Pure copper	Helium	600	3	96.9

Table 5 Effects of gas type, temperature, and pressure on electrical conductivity of CS Cu deposits

4.3.3. Tensile strength & ductility

CS deposits are usually produced in thicknesses ranging from several hundred microns to a few millimeters, and thus unlike other thinner coating processes, cold spray can be deposited in much greater thicknesses to allow for the production of full-size or sub-size tensile coupons per standard test methods like ASTM E-8. Most data on ultimate tensile strength (UTS) and ductility of CS deposits have been extracted from this type of tensile test [87, 217, 227-230].

Some parameters significantly affect the ductility and strength of CS deposits, including particle





velocity [2], gas type [231], gas temperature and pressure [142, 231-233], surface roughness [11, 226], and spray angle [87]. In general, as shown in Fig. 46 [91], if the particle temperature is too low or too high, the particle is too brittle or too ductile, respectively. This illustrates that there is indeed an ideal particle impact velocity and particle impact temperature to achieve optimal mechanical properties in the CS deposit, and that there is a limit to the benefit of increasing temperature in cold spray. Also, gas temperature can affect particles differently depending on their size, since larger particles will retain a higher temperature upon impact [138]. Thus, any variable that can affect particle velocity and temperature can affect the resultant ductility and strength of CS deposits.



Particle impact temperature [°C] Fig. 46 Optimal operating condition in terms of particle impact velocity and temperature (Ref 91)

Huang et al. [216] investigated the effects of gas pressure and temperature on tensile strength of CS Cu deposits and showed that the UTS of CS Cu deposits increased with gas temperature and pressure (Fig. 47). Their study also revealed that the effect of the substrate material on the resultant tensile strength of the CS deposits, which was explained by imposing various levels of deformation on Cu particles with Al alloy and Cu substrates.



In addition, carrier gas type has been shown to have a great impact on tensile strength and ductility of CS deposits due to the velocity differences of the impacting particles. As shown in Fig. 48, N₂ and He influenced the mechanical properties of Ti-6Al-4V deposits differently, *i.e.*, He sprayed deposits showed greater strength and ductility than those of N₂ sprayed deposits (curves #2 & #4, respectively). Similar trends have been reported for other CS deposited materials [5, 11, 18, 80, 234-236]. Fig. 48 also illustrates the difference between the mechanical properties of the CS deposits and the corresponding bulk materials (curve #1). Efforts have been made to improve these properties to that of the bulk materials. One method has been heat treating deposits, which will be discussed later.

Schmidt et al. [237] have also shown that significant increases in the ductility of CS deposits is possible using an optimized nozzle design. They reported increases in tensile strengths of Cu deposits from 50 MPa to as high as 250 MPa with nozzle design improvements and with HPCS systems.

4.3.4. Conductivity

In the CS process, the depositing particles do not melt, and thus the deposits getter less oxygen compared to conventional thermal spray deposits. These features



Fig. 47 Tensile strength of Cu deposits processed with N2 as the carrier gas and gas pressure of: (a) 3 MPa, and (b) 4 MPa (Ref 214)



Fig. 48 Typical stress-strain curves for Ti-6Al-4V substrate and deposits in different conditions (Ref 232)





enable the production of high conductivity Cu deposits with low porosity, low oxide content, and low-thermal stresses [5, 25, 129]. The deposit undergoes SPD and subsequent phenomena, such as dynamic recrystallization and partial chemical reaction [2, 18, 32, 39, 238]. Thus, in combination with low oxygen content, the thermal conductivity is mainly affected by the high dislocation density and the low mean free path of electrons.

According the International Annealed Copper Standard (IACS), the conductivity of oxygen free bulk Cu at room temperature (RT=20°C) is γ =57.14 m/(Ω cm), which corresponds to a resistivity of q =1.7 Ω V cm [135]. As shown in Fig. 49, CS Cu deposits using N₂ process gas exhibited 63% of the bulk conductivity. However, HVOF and arc-sprayed coatings reached only 39% and 19% of the bulk conductivity, respectively. The highest reported conductivity (Coddet et al. [239]), which was 97% of bulk Cu conductivity, was achieved for pure Cu deposits.



Fig. 49 Conductivity of Cu deposits processed by CS, HVOF, and arc spraying in the as-deposited state and after different annealing conditions. Annealed bulk Cu serves as reference material (Ref 134)

The electrical conductivity of CS deposits changes with process parameters, including gas temperature and pressure, particle velocity, and carrier gas type. For example, higher gas pressure increases electrical conductivity, as shown in Fig. 50(a) [5, 135, 240, 241]. This figure also shows that increasing the gas temperature can increase the electrical conductivity as a result of reduction in porosity and superior bonding. Increasing particle velocity has a similar effect (Fig. 50(b)),





especially when it is within the deposition window. Watanabe et al. [242] showed that CS Cu deposits sprayed with N_2 (has temperature of 800°C and pressure of 3MPa) exhibited an electrical resistivity of 1.9 – much greater than that of the bulk Cu. They attributed this increase in conductivity to an optimal micro-texture obtained during deposition.



Fig. 50 Variations in electrical conductivity of (a) silver deposits by carrier gas pressure (Ref 237) and (b) Cu deposits by particle velocity (Ref 238)

CS systems can also affect electrical properties of CS deposits [76, 238]. The high plastic deformation upon impact, particle flattening, and related high quality of the deposit via HPCS led to electrical conductivities up to 79 IACS in the as-deposited condition, compared with only 46 IACS obtained with LPCS.

Overall, electrical conductivity is linked to the bond quality between deposited layers and to





porosity level. Table 6 summarizes the effects of different parameters reviewed in this section (carrier gas type, temperature, and pressure) on electrical conductivity of CS Cu deposits.

Table 6 Flexural strength and strain results for the sintered CS samples (Ref 282)				
	Flexural strength, MPa	Flexural strain		
As-deposited	187 ± 41	0.026 ± 0.022		
Sintered 1250-60 min	1225 ± 122	0.183 ± 0.037		
Sintered 1250-180 min	1512 ± 207	0.161 ± 0.025		
Sintered 1250-60 min + aged	1475 ± 130	0.160 ± 0.046		
Sintered 1250-180 min + aged	1651 ± 658	0.159 ± 0.045		

4.3.5. Residual Stress

Deposit integrity may be loosely defined as the quality of bonding between particles within the deposit, and between the deposit and the substrate. Deposit integrity is also influenced by residual stresses locked within deposits. As is the case in all thermal spray deposits, residual stresses can lead to peeling and delamination of the deposit [5, 17]. As a result, understanding, prediction and control of internal stress accumulation can contribute to improved CS deposit performance.

A common presumption is that CS deposits have compressive residual stresses. There have been experimental [11, 12, 113, 221, 243-248] and numerical [8, 12, 213] studies of residual stress, which confirm this view. Examples of measured residual stress profiles for different combination of CS deposits and substrates are shown in Fig. 51. The plots show that the sign, magnitude and profile of residual stresses in CS deposits depend on various CS parameters, especially on the hardness of the depositing powder and substrate material. Residual stress relaxation at the interface between the substrate and the deposited material can also be observed in some cases. The residual stress is also greater when the depositing material and the substrate have dissimilar hardness (Fig. 51(e)).





Fig. 51 Measurement of residual stress distributions through thickness for (a) Cu/Cu, (b) Cu/Al, (c) Al/Cu, (d) Al/Al (Ref 244), (e) 5052/7075Al and 5052/Al (Ref 112), and (f) Mg/ Al (Ref 242)

Note that after spraying, the substrate-deposit system is normally cooled to room temperature, so that a thermal misfit strain may arise. Thus, the residual stress can be influenced by the heat input, spraying kinematics, and the associated thermal history. Based on these factors, in addition to substrate and deposit material properties and dimensions, the mean residual stress in the deposit can be compressive or tensile [246-248].

4.3.6. Fatigue

Fatigue strength of CS materials is a major concern, since CS repair parts and components are often used in applications involving cyclic loads. Reports about the influence of cold spraying on fatigue strength is scarce. Furthermore, testing procedures vary widely, and the interpretation of





results is generally complex. Not surprisingly, fatigue results are sometimes contradictory. Thus, we choose to report only results from the most thorough studies here.



Fig. 52 Mean number of cycles prior to failure for: (a) the bare, Alclad, and CGDS Al-Co-Ce deposits on 2024-T3 (Ref 11), and (b) pure Al and Al7075 deposits on Al 5052 (Ref 247)

Multiple factors affect fatigue strength of metallic CS deposits. Both spraying parameters and material properties strongly affect the residual stress in cold sprayed materials and thereby, the fatigue strength [11, 115, 249-251]. Ghelichi et al. [11] reported greater fatigue endurance in Al5052 coated with Al7075 compared to specimens deposited with pure Al, as shown in Fig. 52(a). They concluded that the fatigue strength of the treated specimens followed the fatigue endurance of the stronger of the two materials, *i.e.*, the grit blasted substrate in the case of pure Al on Al5052, where the deposited material in the case of Al7075 on Al5052.

In the same study, the residual stresses induced by the deposit on the substrate played a minor role in fatigue endurance increase compared to the contribution of the type of the deposited material and coating parameters. However, Sansoucy et al. [251] studied the fatigue strength of Al-Co-Ce





deposits (Fig. 52(b)), and attributed the superior fatigue life of the deposits to inhibition of crack propagation in the deposit by the residual compressive stresses in the deposit. The presence of compressive residual stress, which is generally generated in CS deposits during deposition, can have beneficial effects in retarding crack propagation under fatigue loading [11].



Fig. 53 Effect of surface preparation on fatigue life of: (a) Al/2024-T351 (Ref 220) and (b) Al7075/Al5052 deposits (Ref 11) Surface preparation greatly influences the fatigue properties of CS deposits. The standard surface preparation for the CS process involves shot peening or grit blasting, both intended to impart nearsurface compressive stresses and increase the surface roughness of the substrate. Ziemian et al. [221] investigated the effect of surface preparation for Al2024-T351 substrates before depositing pure Al. As shown in Fig. 53(a), their results indicated that the fatigue strength increased up to 50% with CS Al deposits. They attributed this increase to the work hardening induced by grit blasting and particle impacts. Ghelichi et al. [11] confirmed this rationale for increased fatigue resistance of 5052 substrates deposited with CS Al7075 deposits (Fig. 53(b)). In their study, the fatigue strength of the coated series increased 13% and 20% with respect to grit blasted samples.





As shown in Fig. 53(b), they also investigated the effect of powder state, *i.e.*, microcrystalline vs. cryomilled NC, by depositing cryomilled NC Al7075 onto Al5052 substrates. The authors highlighted that using cryomilled NC powder slightly increased the fatigue life with respect to the materials deposited using microcrystalline powders [11].

4.3.6. Corrosion

Published results have demonstrated the potential of the CS process for refurbishment of metallic parts and components, especially for corrosion protection. This potential derives primarily from the fact that CS deposited materials are highly consolidated (high density) and impermeable (without interconnected porosity) microstructures. These features are critical to the high corrosion resistance of CS deposits [252, 253].



CS of many materials with potentially high corrosion resistance, such as Zn, Al, Ni, Ta, Ti, SS, $Al+Al_2O_3$, Al-Mg, and brass, has been explored for biomedical, medical, marine, hot, and high pressure applications. In all these applications, the deposited CS material acts as: 1- a anodic layer with sacrificial protection for the substrate [5, 16, 142, 252-263], 2 – a passive oxide layer [231, 264-266], or 3 - a dense cathodic or corrosion resistant protective layer [142, 269-274].





For instance, Blose et al. [258] reported enhanced corrosion protection of steel substrates with CS Zn, Al, and Zn-Al coatings against wet corrosion. Karthikeyan et al. [257] showed that corrosion resistance of the CS Al deposits was greater than that of bulk Al. Fig. 54 shows the polarization behavior of a CS Al coating and Al bulk. Passivation of the coatings is first linear, then slightly non-linear, followed by linear behavior again. This behavior indicates re-passivation of the coatings [254].



Fig. 55 (a) Open-cell potentials of HPCS Ta and Cu coatings and Fe52 substrate as a function of exposure time in 3.5%NaCl solution (Ref 264), (b) Polarization behavior of CS Ta coatings (CSTa1 on Al, CSTa2 on Cu and CSTa3 on steel), Ta bulk material in 1M KOH (Ref 270)

HPCS deposits have yielded excellent corrosion properties for a variety of of alloys, such as Cu [238, 270], Cu alumina [276], Ta [270, 275], Ti [277], Ti+HAP [278], Ni-Cu [269], Ni-Cr [274], SS and SS mixed with Co-Cr [142], and WC-Co [279]. Fig. 55(a) shows dense coating structures of the HPCS Cu and Ta coatings, having open-cell potential behavior similar or superior to corresponding bulk materials. Bulk Ta and dense HPCS Ta coating passivate rapidly, and above





the passivation potential, the corrosion rate drops to a negligible value in most environments due to the stable passive layer [275]. Bulk Ta and cold-sprayed Ta coatings on different substrates (CS-Ta1 on Al, CS-Ta2 on Cu and CS-Ta3 on steel) exhibit similar anodic corrosion behavior (Fig. 55(b)), which testifies to the high density and impermeability of the CS Ta coatings. The polarization behavior of inert-plasma-sprayed Ta coatings deposited on Fe52 steel substrates, shown as IPS Ta1 in Fig. 55(b), were also analyzed in this study, which indicates these coatings probably contained through-porosity, allowing the electrolyte to reach the substrate and prevented the formation of a uniform passive film.






Fig. 56 Polarization behavior of CS SS + CoCr coatings in (a) as-deposited and (b) annealed (1100 C) coatings (Ref 141) Post-deposition heat treatments can be employed to enhance cold spray deposits. For example, Al-Mangour et al. [142] measured corrosion properties of a coating produced by CS of blended powder of SS 316L and Co-Cr alloy L605 (L605 exhibits corrosion resistance superior to 316L but is difficult to process). The authors showed that 25% Co and 33.3% Co coatings had higher corrosion potential (i.e. higher corrosion resistance) than that of pure SS, as shown in Fig. 56(a). However, in the 50–50% coatings, the corrosion rate was much higher than the pure SS because of





much higher porosity in this coating compared to 25% and 33% coatings (about 4.5% compared to about 0.9–1.5%). After annealing at 1100°C, the corrosion potential increases for all coatings (Fig. 56(b)), especially for the 50–50% coatings, shown by arrow in Fig. 56(a). Similar results were reported by Chavan et al. [16], who showed that a heat-treated Zn coating had reduced corrosion current density, indicating corrosion protection superior to the as-deposited materials. The improved corrosion resistance derived from the microstructural modifications of the deposit which occurred during an annealing HT, such as those described in section 3.4.

4.4. Effect of post CS heat treatments on mechanical properties

The effects of heat treatment (HT) on the mechanical properties of cold-worked material are summarized in Fig. 57 [280], and CS deposits are expected to behave similarly. Generally, residual stresses start to decrease during recovery and are completely relaxed during recrystallization. Furthermore, the largest differences between properties occur after recrystallization. Hardness and strength decrease, while ductility increases. New grains are generated by recrystallization, and the new grains grow during the grain growth phase if left at elevated temperatures. Hardness and strength are relatively stable in the grain growth region, while ductility continues to increase [280]. The effects of post-process HTs on mechanical properties of CS deposited materials are reviewed in this section.



Fig. 57 Cold-worked structure and properties. Effects of annealing temperature on recovery, recrystallization, and grain growth (Ref 275)

4.4.1. Post-HT hardness

Post-CS annealing can significantly affect the hardness of CS deposits [280-281]. Hardness behavior as a function of annealing temperature is shown in Fig. 58. The microhardness of the as-deposited and annealed 304 SS coatings is shown in Fig. 58(a) [281]. Annealing leads to a decrease in the microhardness of the deposit. The minimum microhardness of the annealed deposit was obtained at 950°C (~201 HV), a value comparable to that of an annealed bulk 304 SS. The softening is caused by dislocation annihilation, relaxation of peening stress and cold working, and grain growth. Similar trends have been reported for other CS deposits [5, 18, 135, 141, 208, 240, 282].

However, in some cases, the hardness of deposited materials increases with annealing temperature [214, 220, 283]. Li et al. [283] measured microhardness values of CS Ti deposits, and reported that average values of microhardness gradually increased as the annealing temperature increased from 850°C to 950°C (Fig. 58(b)). This behavior was attributed to further consolidation and densification of the deposits during post-spray HT (such as annealing) leading to closure of pores, inter-splat boundaries, and cracks in the microstructure [214], especially when N₂ is used as the carrier gas [223].



Fig. 58 Microhardness variation as a function of HT temperature for: (a) SS 304 (Ref 276) and (b) Ti deposits (Ref 277)

4.4.2. Post-HT bond strength

To attain greater bond strength and promote coating performance, CS deposits can be heat treated. Stoltenhoff et al. [135] studied the effect of annealing of CS Cu deposits on different substrate materials, *i.e.*, Cu/Al, Cu/Cu, and Cu/low carbon steel. As shown in Fig. 59, post-CS HTs strongly influenced the adhesion strength of Cu deposits to different substrate materials. This figure also shows that the increase in bond strength correlates with the annealing temperature. In Al/Cu case, post-CS annealing (1h at 400°C) increased the bond strengths from ~ 40 to ~ 60 MPa. The adhesion of CS deposits on low carbon steel substrates did not exceed 10 MPa in the as-deposited state, which is similar to the values achieved by arc or flame spraying, but less than those obtained by using HVOF spraying (25 – 30 MPa). Note that an almost linear trend of increase with annealing temperature, which is observed for "soft" substrate materials (e.g., Al or Cu) cannot





be confirmed for deposits sprayed onto steel substrates. Hussain et al. [284] also investigated the adhesive strength of Cu deposits on Al substrates and reported an increase in the bond strength after HT at 400°C (adhesion strength increased from 57 MPa to 69 MPa).



Fig. 59 Bond strength of CS Cu deposits on various substrate materials (Al: aluminum, Cu: copper, St: low carbon steel); asdeposited and annealed states (Ref 134)

However, in some cases an opposite trend has been observed, *i.e.*, a decrease in bond strength with increasing annealing temperature. As shown in Fig. 60(a), Al deposits on Mg substrates (black line) show a decrease in bond strength after annealing, and the interfacial bond becomes weaker with increasing annealing time. This finding was attributed to formation of Al_xMg_y IMs at the interface [194]. According to Fig. 60(b), the thickness of IM compounds increased with holding time at the anneal temperature [191]. Because of the brittleness of such IMs, they adversely affect the deposit bond strength.





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Fig. 60 (a) Variation of the shear bond strength of different deposit/substrate systems before and after post-spray annealing treatment (Ref 192), and (b) Thickness measurement of the IM layers vs. holding time of the HT at 400 C under vacuum (Ref 189)

4.4.3. Post-HT tensile strength & ductility

One of the biggest mechanical challenges for CS arises from the reduction or loss of ductility in the deposit because of the high degree of cold work caused by the plastic deformation of the impacting particle. However, ductility can generally be improved with post-CS HT, as shown in Fig. 61 [31]. For example, Ogawa et al. [26] showed that post-CS annealing of CS Al specimens at temperatures as low as 270° C restored ductility compared to untreated specimens. Under tensile loading, the strength and ductility of the heat-treated specimens was $2\times$ and $5\times$ greater than that of the untreated specimens, respectively.





Similar behavior deposits were reported for CS 7075Al [182]. As shown in Fig. 61, the asdeposited (AD) 7075 material showed the lowest UTS and ductility among all the conditions, again as a consequence of the extensive cold work introduced into the microstructure during CS. The cold work in the AD material produces dislocation structures in the particles, and UFG structures at the PPBs, as shown previously in Figs. 24 & 28.



Fig. 61 (a) Ultimate strength and (b) the elongation of asdeposited and annealed deposits at different temperatures (Ref 180) The mechanical properties of the CS deposit increased after all of the post-CS HTs [182]. Fig. 61(a&b) shows that low-temperature HTs (T6, T7X, T73 and stress relief (SR)) increased the UTS and ductility of the CS 7075 Al deposit. The increase in the UTS in these conditions was attributed to precipitation of the IM phases η' and η . The precipitates impede dislocation movement and thereby increase strength. In all of the low-temperature HT conditions, the hardening effect by





aging dominates the softening effects by recovery, and as a result, samples heat-treated at lower temperature are stronger and more ductile.

Heat treatments at higher temperatures (annealing and solute solutionizing (SS)) also improved strength and ductility of CS 7075 Al deposits, as shown in Fig. 61. The graphs show that annealing of the CS deposits led to increases in both UTS values and ductility, which increased from \sim 3.2% in the AD condition to \sim 14% and 10% in annealed and SS+T6 conditions, respectively.

Mechanical property enhancement occurs because of three main reasons. First, annealing induces diffusion and microstructural sintering through the particle interface and promotes metallurgical bonding and mechanical interlocking [31, 182, 183, 285]. Second, annealing above optimal temperatures leads to grain growth, increasing the interfaces between particles and resulting in reduction of voids and un-joint interfaces [5, 18]. Third, in CS materials, the stored energy as plastic deformation can drive recrystallization and consolidation when the material is heated [26, 230, 280]. In certain cases, strengthening can also occur due to the precipitation of strengthening phases, which can also be smaller in size and with a uniform distribution due to their finer distribution from powder particle grains.

If CS deposits are fully consolidated before post processing, post-CS annealing decreases the flow stress and increases ductility [284-288]. This behavior has been rationalized by considering the microstructural recovery of the hardened as-deposited splats, diverting plastic strain from the interfaces during indentation, as well as grain growth. Choi et al. [62] supported this hypothesis by producing CS Al specimens, and subsequently annealing them at 300 °C for 22 h in argon or laboratory air. They found that annealing softened CS Al enough to promote ductile behavior, but the flow stresses decreased in the CS deposits compared to the bulk material. Similar behavior can be observed for 1100°C in Fig. 62(a) and all temperatures in Fig. 62(b).







Fig. 62 Stress-strain curves of (a) 316L (Ref 140) and (b) Cu (Ref 221) in as-deposited and different annealed conditions Nevertheless, CS deposits processed with nitrogen show brittle failure under relatively low tensile stress, even after thermal annealing. Only the closure of PBBs which are just under compressive contact was observed and the highest obtained elongation was ~8 % [1, 5, 18].

CS of nickel-based superalloys (*i.e.*, Inconel 718) has been challenging mainly because of its high strength and the occurrence of nozzle clogging. This problem has been solved by using HPCS and a non-clogging nozzle material combined with a nozzle water cooling system, enabling practitioners to deposit Inconel 718 [289, 290] and study the mechanical properties of as-deposited and heat treated coatings. For example, Wong et al. [290] showed that the as-deposited and heat-treated deposits produced at high particle velocity (787 m/s) were slightly denser (1.9-2.7% porosity) than those produced using lower velocity (741 m/s) conditions (3.2-3.8% porosity). They





also reported that after heat treatment at 1250°C for 1h, Inconel 718 deposits showed enhancement of metallurgical bonds at PPBs, which increased the mechanical properties. Thus, the UTS and ductility reached up to 763.6 MPa (62% of the corresponding bulk material) and 24.7%, respectively (Fig. 63).



Fig. 63 Tensile test results for as-deposited and heat-treated Inconel 718 coatings produced with a particle velocity of (a) 741 m/s and (b) 787 m/s (Ref 283)

The difficulty in achieving metallurgical bonding in CS Inconel 718 stems from insufficient deformation and motivated investigations of post-CS sintering as a means of enhancing mechanical properties. Levasseur et al. [289] confirmed that upon sintering and conventional aging, the flexural strength of CS Inconel 718 deposits increases from 187 to 1651 MPa, and ductility from 0.026 to 0.183, as shown in Table 7.





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Condition	Flexural strength, MPa	Flexural strain
As-deposited	187 ± 41	0.026 ± 0.022
Sintered 1250-60 min	1225 ± 122	0.183 ± 0.037
Sintered 1250-180 min	1512 ± 207	0.161 ± 0.025
Sintered 1250-60 min + aged	1475 ± 130	0.160 ± 0.046
Sintered 1250-180 min + aged	1651 ± 658	0.159 ± 0.045

Table 7 Flexural strength and strain results for the sintered CS samples (Ref 221)

4.4.4. Post-HT conductivity

Low oxide content and high density of CS deposits generally enhance the conductive properties of such coatings. Fig. 64 shows how these features increase the conductivity of CS Cu deposits compared with other thermal spray processes. This figure also shows that post-CS annealing increases electrical conductivity of Cu deposits, and the higher the annealing temperature, the greater the electrical conductivity.



Fig. 64 Conductivity of CS Cu deposits, HVOF spraying and arc spraying in the as-deposited and after different annealing conditions. Annealed bulk Cu serves as reference material (Ref 134)

The effects of post-CS annealing on thermal and electrical conductivities of CS alloys and composites are reported to be strictly related to microstructural changes in CS deposits during annealing. During annealing, recovery and recrystallization reduce defects such as porosity, dislocations, interstitials, vacancies, non-well-adhered splat interfaces, and GBs, all of which reduce thermal and electrical conductivities by scattering electrons or photons [5, 18, 230, 280].





Post-CS HTs also promote bonding between deposited particles, which enhances the coating quality and consequently, increases the conductivity of CS deposits [238, 291, 292].

The microstructural anisotropy in CS deposits can also influence electrical resistivity, as demonstrated by Li et al. [208]. However, extensive recrystallization during annealing process, especially at PPB's, leads to equiaxed grain structures in CS deposits. This phenomenon improves the bonding between PPB's and increases electrical conductivity in both directions (see Fig. 65).



Fig. 65 Annealing effect on the electrical resistivity of CS Cu deposit in the parallel and perpendicular directions (Ref 206) This relationship also provides a method to use conductivity to correlate with increased mechanical properties as a non-destructive inspection method, since the same factors which also improve conductivity are general good for mechanical properties as well.

However, post-CS annealing above the optimal temperature is not recommended. Soe et al. [292] studied the effect of post-CS annealing on thermal conductivity of Cu deposits and concluded that there is an optimal temperature up to which thermal conductivity increases with annealing temperature. Above this temperature, however, voids rearrange and concentrate along GBs, which can degrade mechanical, physical, or other characteristics of the CS deposits. Fig. 66 summarizes the relation between annealing condition and thermal conductivity.



Annealing environment can affect the extent to which the conductivity of CS deposits can be enhanced [291]. Fig. 67 indicates that annealing Cu deposits in vacuum at 300°C is sufficient to obtain conductivity values similar to that of bulk Cu. On the other hand, even a 4 hour anneal at 450°C in air did not yield the required conductivity.

4.5. Key messages

CS produces severe plastic deformation on the impacting particles, and the non-uniform nature of the deformation causes an inhomogeneous microstructure in the deposits. Micro-scale mechanical property evaluations show the presence of UFG structures in PPB's and greater hardness compared with particle interiors. Bulk-scale mechanical properties vary with process parameters used during cold spray, including particle velocity, gas temperature and pressure, substrate material, and spray angle. The ductility of CS deposits is usually low, but can be recovered to near-bulk level by post-CS annealing.



Fig. 66 Schematic description on optimization of annealing process; (a) grains in as-deposited Cu coating, (b) uniformly grown grains after optimal annealing, and (c) abnormally grown grains over optimal annealing temperature (Ref 285)

If the post-CS HT is optimal, the deposit quality can be enhanced, particularly the bond strength, by increasing bonding area between deposited particles and reducing porosity in the microstructure. For the same reasons, the conductivity of CS deposits can match or exceed the conductivity of bulk materials. The compressive residual stress in CS deposits increases fatigue





resistance. Thus, annealing temperatures should be kept as low as possible to avoid redistribution of residual stress or even part distortion. Low-temperature anneals also help preserve the fine microstructure and superior coating hardness and their related benefits in the as-deposited materials. Moreover, even if beneficial, a post-deposition anneal can be unacceptable from an industrial perspective due to the component size or base material or even specific production process.



Fig. 67 Effect of annealing environment and temperature on conductivity of CS Cu deposits (Ref 284)

5. Perspectives

CS is a versatile process for repair and refurbishment of metallic parts and components, and is also useful for the manufacture of new parts. The particle deformation behavior and process parameters determine both particle velocity and temperature, and these factors strongly influence the microstructure and mechanical properties of the deposit. Understanding these relationships will help researchers and practitioners to design spraying procedures that optimize mechanical properties (hardness, strength, ductility, or fracture toughness) of the deposits for the intended application. The selection of procedures includes choice of the cold spray system (to achieve suitable particle velocities), particle size, and temper of the feedstock powder (which affects





degree of particle deformation). Current and future efforts to optimize CS systems should focus on improving control of the particle flow and preprocessing the feedstock powder to increase deformation and bonding during deposition as-well as very appropriate taking advantage of the ability to perform post-CS HTs on as-deposited materials. Such focused effort will accelerate advancement of the technology, increase the number of applications, and reduce process costs.

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