

Erosion-Corrosion Performance Evaluation of Different Materials for Oil Sand Application

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Abstract

Material loss due to erosion-corrosion decreases the throughput and useful life of the equipment. Simultaneous action of erosion and corrosion is responsible for the high degradation of the hydro-transport equipment. To minimize the operational cost, based on the degradation severity in a specific operation, different classes of materials are being used for different applications. Pipeline steels (i.e. plain carbon, API steels), for example, are relatively cheap and do not provide good wear resistance. These types of steels are typically used for less severe applications such as tailings. Chrome white irons (CWIs) and WC-based overlays, on the other hand, are usually used for more severe wear conditions (i.e. hydro-transport pipeline). In this study, we have evaluated the erosion-corrosion performance of 4 homogeneous materials (pipeline and abrasion-resistant steels) and 5 materials containing different types and amounts of carbide (chrome white iron and WC-based overlay). Erosion-corrosion test was performed inside a slurry pot at 45°C, in an aqueous slurry containing 35wt% natural silica sand and 3.5wt% NaCl. The wear performance of these materials was evaluated based on the total erosion-corrosion (E-C) rate as well as the separate components of synergistic effect. In the current test condition, WC-based overlays demonstrate the best erosion-corrosion resistance. For all carbide-containing materials, it was found that matrix wear influences the extent of carbide degradation. For the carbides to provide good erosion-corrosion resistance, the surrounding matrix that supports the carbide should have sufficient wear and corrosion resistance. Dominant wear mechanisms for homogeneous and carbide-containing materials have also been identified.

Key words: Erosion-corrosion, synergistic effect, WC-overlays, Chrome white irons, Abrasion-resistant (AR) steel, Pipeline steel, slurry pot,

1 Introduction

Erosion-corrosion plays an important role in determining the durability of the erosive/corrosive slurry handling equipment. The combined action of erosion and corrosion produces a high degradation rate. Material loss rate due to erosion-corrosion is usually significantly higher than the sum of material loss rate due to pure erosion and pure corrosion acting separately [1]–[4]. The degradation in erosion-corrosion is primarily by matrix extrusion due to abrasive particle impact. Corrosion processes act to accentuate the damage. Previous investigation showed that erosion-corrosion synergy can contribute more than 50% to the total material loss rate [5]. The selection of appropriate material (i.e. capable of resisting mechanical and electrochemical synergistic degradation) is crucial to reduce the overall maintenance cost and extend the service life of the equipment [6]–[9]. However, concurrent erosion and corrosion interactions are complex and, make it is difficult to determine the material loss rate with sufficient accuracy for reliable equipment lifetime prediction.

To improve both performance and transportation efficiency, the steel pipe industry has developed different grades (i.e. Grade A to API X120) of steel. Steel microstructure is a key variable in determining the mechanical properties of plain carbon and high strength low alloy (HSLA) steel.

In the case of HSLA, grain refinement is one of the main methods by which both strength and toughness can simultaneously be improved. Most of the HSLA steels undergo thermo-mechanical rolling with or without accelerated cooling. Higher steel grade provides superior mechanical and metallurgical properties. However, superior mechanical properties do not always translate to superior wear resistance [10]–[13]. Abrasion resistant (AR) steels are generally used for applications requiring low-stress abrasion resistance. Limited work has been done to evaluate the erosion-corrosion performance of these steels. Due to its low alloy content, tribo-corrosion performance of these steels is not sufficient for some engineering applications. Few studies have been focused on surface treatment to overcome this drawback. Ilyas has studied the effect of plasma nitriding treatment on tribo-corrosion properties of different types AR steels [14]. It was found that nitriding changed the electrochemical degradation mechanism of the surface to pitting corrosion and significantly improved the corrosion and wear resistance of AR steels.

High-Cr white irons are extensively been used in mining and mineral processing industries to minimize the erosion-corrosion damage of critical areas in the slurry handling equipment (*i.e.* slurry pumps) [15]–[26]. Wear and mechanical properties of high-Cr white irons are derived from a well-controlled matrix microstructure and M_7C_3 carbides (M represents metallic species, most commonly Fe and Cr). In the Fe–Cr–C system, the type and form of eutectic carbides change with an increase in Cr content. In unalloyed white cast irons (<5% Cr and up to 2% C), continuous M_3C type carbides are formed during eutectic solidification. At 8–10% Cr, the eutectic carbides (M_3C) become less continuous, and above 10% Cr, the eutectic carbides changes to the discontinuous M_7C_3 type [20], [27]–[31]. The high abrasion and erosion resistance of Cr white cast irons is derived mainly from the composite microstructure where the Cr-carbide is harder than the most common abrasives in mining. These carbides can protect the matrix from severe wear. Meanwhile, the matrix supports the carbide to minimize microchipping. An ideal microstructure for optimal wear resistance should contain a high volume fraction of hard and tough carbide and a strong/hard matrix that can withstand the wear attacks under given situations. Hence, improving the wear performance of white irons relies heavily on optimizing the chemistry and physical metallurgy of the matrix [32].

Erosion-corrosion in the slurry handling equipment can also be minimized by employing protective overlays [33]–[35]. The erosion resistance of weld overlays is achieved due to the combination of matrix toughness and hardness of the reinforcing particles [36]–[38]. It also depends on the volume fraction, size, and distribution of the hard phases [36], [39], [40]. Erosion-corrosion is a complex phenomenon and erosion-corrosion resistance not only depends on the reinforcing particles, but also the matrix composition and carbide matrix interface [39], [41]–[44]. The wear resistance can be improved by increasing the volume fraction of carbides [38], [45]. Tungsten carbide (WC) has widely been used as the hard material for protective overlays. High hardness, a low coefficient of thermal expansion, a certain amount of plasticity, and good wettability by molten metals make it a suitable candidate for overlay. The main disadvantage of tungsten carbide is its low heat of formation. As a result, carbides can easily be dissolved by molten metals which may affect the wear performance. WC-based overlays in this study were applied via a vacuum furnace braze infiltration process, resulting in much lower dilution. The relatively ductile matrix, such as chromium and nickel based alloy can greatly improve both the toughness and corrosion resistance of the overlays.

Ball concluded that, erosion in homogeneous materials can be regarded as a two-stage process, *i.e.* accumulation of strain preceding the loss of material by micro-fracture [46]. Hence, materials that can absorb strain and work harden without the initiation/ propagation of fracture, will provide good erosion resistance. Matrix hardness also plays an important role in determining wear resistance. Material wear performance mostly depends on the ratio of the abrasive hardness (H_a) to the worn surface hardness (H_m) [47]–[49]. If the abrasive is softer than the wear surface (*i.e.*

$H_a/H_m < 1.0$), the wear rate will be minimal. For materials containing hard particles, the effect of the matrix is related to the degree of protection offered by the carbides [50]. If the carbides protect the matrix from the abrasive particles, the role of the matrix is primarily to provide mechanical support. However, if the matrix is preferentially removed by the abrasive, the carbides may become unsupported and susceptible to spalling and fracture. In this circumstance, the abrasion resistance of the matrix is critical, as this controls the rate at which the carbides become unsupported and fractured [51]. However, the role of the matrix is more complex during erosion-corrosion. It was found that preferential matrix wear creates voids at the carbide/matrix interface and accelerates the carbide pull out [45]. Hence, to provide mechanical support, the matrix not only needs protection from the abrasive particle impact but also requires good corrosion resistance.

The objective of this study is to evaluate the tribo-corrosion behavior of different classes of materials. This study also focuses on the effect of microstructure and matrix hardness on the wear performance for homogeneous materials (*i.e.* pipeline and abrasion resistant steel). For materials containing hard particles (*i.e.* white irons and WC-based overlays), the effect of carbide size, shape and volume fraction on tribo-corrosion performance has also been evaluated. In this study, erosion-corrosion resistance of white irons was assessed at 45 °C, in an aqueous slurry containing 35 wt% natural angular sand and 3.5 wt% NaCl.

2 Experimental

2.1 Materials characterization

Nine materials from 4 different material classes have been selected for evaluation. This includes 4 homogeneous materials and 5 materials containing different types and amounts of carbides. Homogeneous materials are plain carbon and high strength low alloy (HSLA) from pipeline steel class and 2 abrasion resistant (AR) steels containing low and high amount carbon. Carbide containing materials are white irons (*i.e.* hypo-eutectic, near-eutectic and hyper-eutectic) and WC-based overlays containing different amounts of carbide loading. Nominal matrix composition is tabulated in Table 1.

Table 1 Evaluated test materials and nominal matrix composition.

Class	Materials	Nominal composition*								
		C	Cr	Ni	Mn	Si	Mo	Co	W	Fe
Pipeline steels	Plain carbon	0.15	0.48		0.85	0.28	0.2	-	-	98.0
	HSLA	0.17	0.2	0.05	1.5	0.25	0.07	-	-	97.8
AR steels	AR steel (low C)	0.26	1.4	1.5	1.6	0.7	0.6	-	-	93.9
	AR steel (high C)	0.47	1.2	2.5	1.4	0.7	0.7	-	-	93.0
White irons	Hypo-eutectic WI	2.75	19.2	0.66	0.86	0.45	1.85	-	-	74.2
	Near-eutectic WI	2.9	26.1	0.35	0.8	0.5	1.9	-	-	67.5
	Hyper-eutectic WI	5	35	-	-	-	-	-	-	60.0
WC overlays**	WC-based (50wt%)	-	18.9	70.2	-	-	-	5.7	3.8	1.3
	WC-based (70wt%)	-	7.3	89.5	-	-	-	1.5	1.8	-

* Chemical compositions determined by EDS

** Tungsten carbide includes cobalt-bonded WC

Figure 1 shows the SEM micrograph of pipeline steel. The microstructure of plain carbon steel consists of pearlite and ferrite (Figure 1 (a)-(b)). The banded microstructure in Figure 1 (a)

indicates that the finishing rolling was done after the transformation. In other words, the finishing rolling temperature was lower than the Ar3 temperature. As a result, ferrite & pearlite phases developed before finishing rolling and elongated along the rolling direction. The microstructure of high strength low alloy (HSLA) consists of polygonal ferrite and bainite. The average grain size in HSLA steel is finer compare to plain carbon steel. The microstructure of abrasion resistant (AR) steels is shown in Figure 2. Both AR steels contain martensitic microstructure. The chemical composition of these steels is similar except AR steel (high C) contains slightly higher carbon and nickel. High carbon in the martensitic matrix can improve the matrix hardness and hence the erosion resistance.

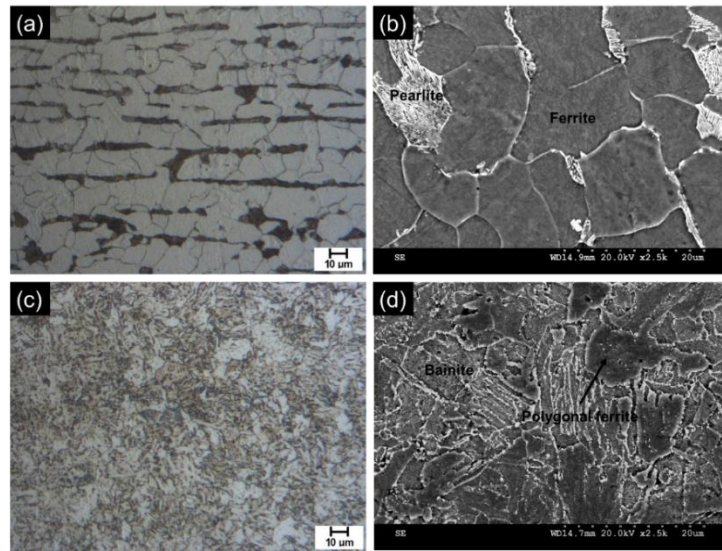


Figure 1 Optical and SEM micrograph of pipeline steels, (a)-(b) plain carbon and (c)-(d) HSLA steel.

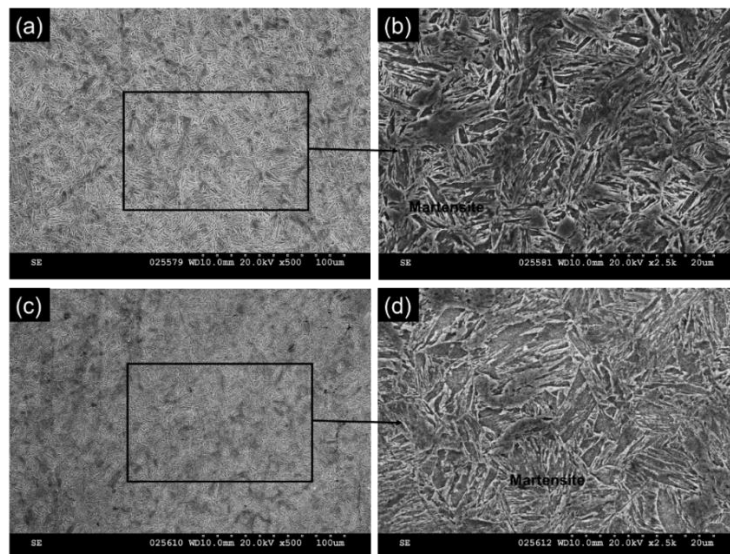


Figure 2 SEM micrograph of (a)-(b) AR steel (low C) and (c)-(d) AR steel (high C).

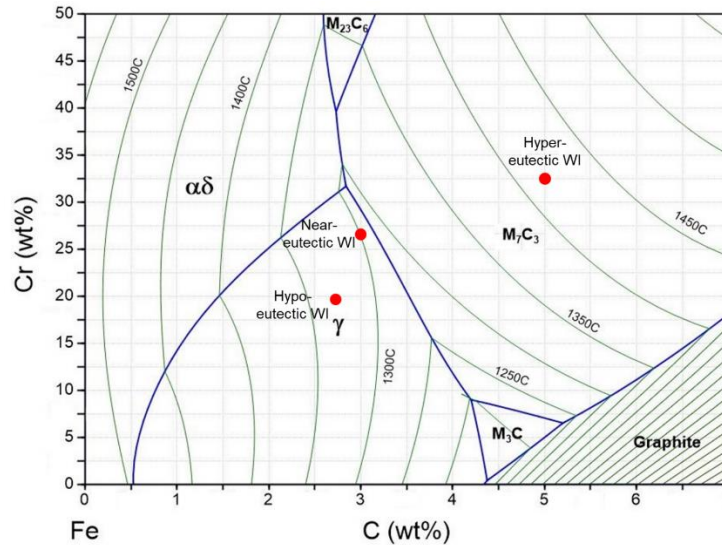


Figure 3 Fe-C-Cr metastable phase diagram illustrating the relative locations of the WI's.

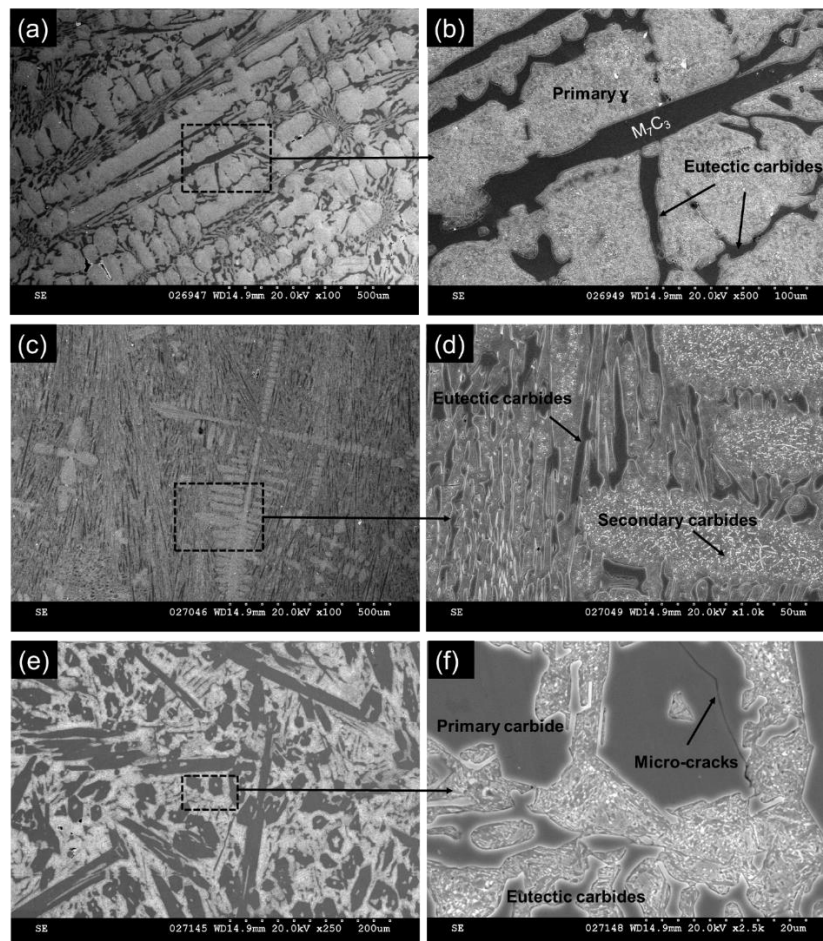


Figure 4 SEM micrograph of (a)-(b) Hypo-eutectic WI, (c)-(d) Near-eutectic WI and (e)-(f) Hyper-eutectic WI.

Relative positions of the evaluated white irons based on their chromium and carbon content are shown in the iron-rich corner of the Fe-Cr-C phase diagram (Figure 3) [52]–[54]. During solidification, the first phase that solidifies in a hypo-eutectic white iron is austenite dendrites (γ). As the solidification progresses, primary austenite dendrites reject carbon and chromium to the liquid [55]. The remaining liquid is pushed towards higher C and Cr contents as it reaches the eutectic valley (right and upward on the liquidus surface in Figure 3). Once the eutectic solidification starts (liquid $\rightarrow \gamma + M_7C_3$), the carbide phase grows spontaneously, thus forming an interconnected eutectic structure in the bulk material [56]. Figure 4 (a)-(b) shows the SEM micrograph of hypo-eutectic white iron. The microstructure consists of primary austenite dendrites in eutectic colonies of M_7C_3 carbides and austenite. Figure 4 (c)-(d) shows the micrograph of near eutectic white iron. The microstructure contains mostly eutectic structures with uniformly distributed fine carbides and a small amount of primary austenitic dendrites. In near-eutectic white iron, carbides are uniformly distributed throughout the structure providing a smaller mean free path. There is also some evidence of secondary carbide precipitation inside the austenite dendrites (Figure 4 (d)). The microstructure of hyper-eutectic white iron consists of primary carbides in eutectic matrix (Figure 4 (e)-(f)). Hyper-eutectic white irons solidify first in the primary M_7C_3 region (Figure 3). As Cr and C-rich carbides are formed, the remaining liquid is rapidly depleted of C and Cr (*i.e.* M_7C_3 requires approximately 8.9%C) as it reaches the eutectic valley (left and downward on the liquidus surface in Figure 3), followed by a eutectic of $\gamma + M_7C_3$.

Figure 5 shows the SEM micrograph of WC-based (50wt%) and WC-based (70wt%) overlay. The majority of the carbides are sintered carbides. Both of the overlays contain a large fraction of fine rounded particles distributed in the matrix. WC-based (50wt%) overlay also contains a mixture of sintered and monocrystalline (blocky) carbides (Figure 5 (a)-(b)). The monocrystalline carbides are smaller than the sintered carbides. Figure 6 shows the EDS mapping for WC-based (70wt%) overlay. Almost all the hard phases are tungsten carbides in a Ni-based matrix. There is also some evidence of Cr-based carbides which formed either independently or around tungsten carbides.

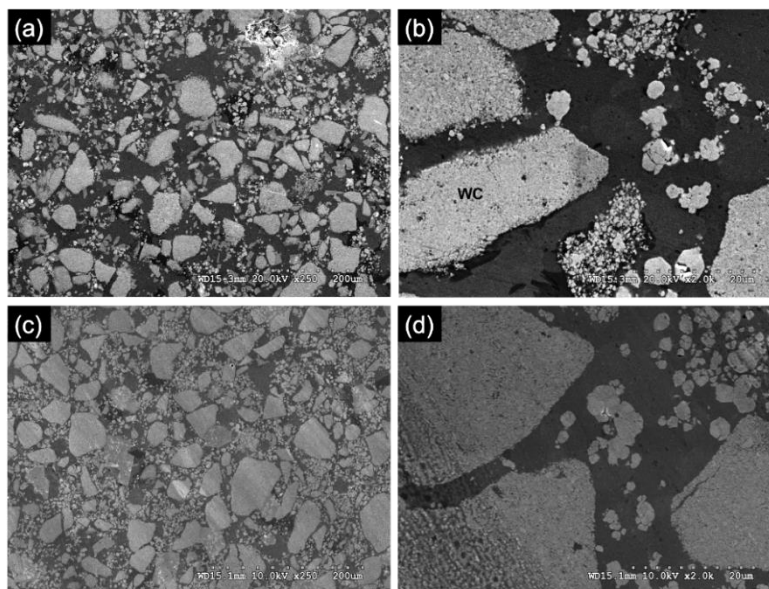


Figure 5 SEM micrograph of WC overlays (a)-(b) WC-based (50wt%) and (c)-(d) WC-based (70wt%).

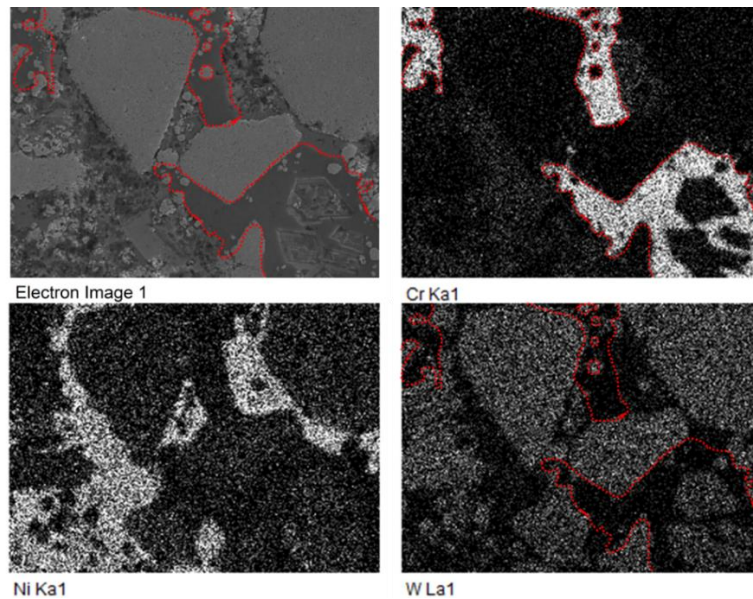


Figure 6 EDS mapping for WC-based (70wt%). Cr carbides are formed either independently or around tungsten carbides.

Average matrix hardness of the tested materials is shown in Figure 7. The hardness of white irons and WC-based materials was measured on the Rockwell C scale. Rockwell B scale was used for pipeline and AR steels. All the hardness was then converted to the Brinell scale (HB) for comparison. Near-eutectic and hyper-eutectic white irons exhibit the highest bulk hardness. Despite a higher carbide fraction, WC-based (70wt%) has slightly lower hardness than WC-based (50wt%) overlay. AR steel (high C) displays relatively higher hardness than AR steel (low C) due to its higher carbon content. Pipeline steel exhibits the lowest hardness among the evaluated materials. The hardness of high strength low alloy (HSLA) steel is slightly higher than plain carbon steel. This might be due to finer grain size and due to the presence of bainite in the microstructure.

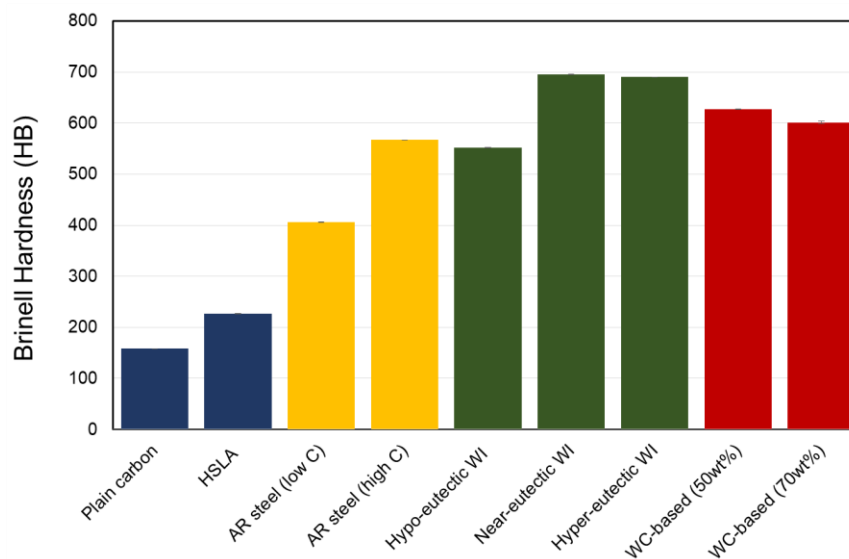


Figure 7 Average matrix hardness of evaluated materials