

# The Great Minds of Carbon Equivalent

## Part III: The Evolution of Carbon Equivalent Equations

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In welding, carbon equivalent (CE) calculations are used to predict heat affected zone (HAZ) hardenability in steels. These CE equations can be used to establish criteria to predict cold-cracking, as they can predict maximum hardness. Dearden and O'Neill believed that the following critical hardness guidelines could be used to judge the cold-cracking tendency of steels [1][2].

- If the HAZ hardness is less than 350HV, no cold-cracking would occur.
- If the HAZ hardness is greater than 400HV, the steel would be prone to cold-cracking.

Dearden and O'Neill realized that the cold cracking tendency wasn't determined only by the steel's chemical composition. They understood that the welding consumable, plate thickness, and weld size would all have an effect. However, the effect of diffusible hydrogen, which can cause cold cracking in the HAZ even when the hardness is lower than the critical 350HV, was not further explored.

In the summer of 1968, Yoshinori Ito and Kiyoshi Bessyo [3] submitted a paper to Commission IX of the International Institute of Welding (IIW) in Warsgawa to propose  $P_{cm}$  as a new carbon equivalent equation:

$$P_{cm} = \frac{C+Si}{30} + \frac{Mn}{20} + \frac{Cu}{20} + \frac{Ni}{60} + \frac{Cr}{20} + \frac{Mo}{15} + \frac{V}{10} + 5B \quad (1)$$

Due to widespread application of the carbon equivalent in Japan, the Japanese Welding Engineering Society (JWES) published its own carbon equivalent equation in 1973:

$$CE_{Wes} = C + \frac{Si}{24} + \frac{Mn}{6} + \frac{Ni}{40} + \frac{Cr}{5} + \frac{Mo}{4} + \frac{V}{14} \quad (2)$$

It should be noted that specification WES-135 of the Japanese Welding Engineering Society lists  $P_{cm}$

as an alternative method of calculating the carbon equivalent of HSLA steels.

In Germany, 1973, Beckert [4] considered that the maximum hardness was related not only to the linear addition of the hardening terms commonly used in carbon equivalent equations, but also to the interaction between carbon and the remaining alloy elements. Hence, he presented a unique carbon equivalent equation (3) and an equation (4) to calculate the maximum hardness, both of which take these interactions into account:

$$CE_{Beckert} = \frac{C+Si}{11} + \frac{Mn}{2.9} + \frac{Ni}{17} + \frac{Cr}{3.2} + \frac{Mo}{3.4} + \frac{Cu}{3.9} \quad (3)$$

$$Hv = (A-B)\exp(-bt)^2 + B \quad (4)$$

Where:

$$B = 167(CE_{Beckert})^{2.42} + 137 \quad (5)$$

$$A = 939C + 284 \quad (6)$$

$$b = \exp(-0.013B + 0.8) \quad (7)$$

$$t = \Delta t_{8/5}$$

The importance of equation (4) is that it was the first to incorporate the cooling rate  $\Delta t_{8/5}$  (the time that a weld cools from 800°C to 500°C) into a hardness equation, providing inspiration for later researchers.

Three years later, Stout [5] developed a slot test to check the weldability of pipeline steels in the field. Like Ito et al. [3] and Beckert [4], he presented a new carbon equivalent equation by considering the effect of Cu based upon his testing results:

$$CE_{stout} = C + \frac{Mn}{6} + \frac{Ni}{20} + \frac{Cr+Mo}{10} + Cu \quad (8)$$

Stout concluded that as long as  $CE_{stout}$  was less

than 0.35, the pipeline steel would not crack. It is interesting to note that this number agrees well with Ito's  $0.35 P_{cm}$ .

One year after Stout's work, Seyffarth [6] was inspired to develop an equation to predict the maximum hardness by combining  $CE_{WES}$ ,  $P_{cm}$ , the interaction between C and Cr, and cooling rate:

$$H_{v-30} = 323.6 - 114.6(\text{Ln}\Delta t_{8/5}) + 11.33(\text{Ln}\Delta t_{8/5})^2 + 123.7(\text{Ln}\Delta t_{8/5})CE_{WES} - 15.88(\text{Ln}\Delta t_{8/5})^2CE_{WES} - 1299C - 79.11Si - 120.7Mn + 79.22Ni - 539Cr + 2830C*Cr + 620.8CE_{WES} + 875.4P_{cm} \quad (9)$$

Owing to their excellent strength and toughness performance, niobium-alloyed large-diameter pipe steels have been popular for transporting oil and gas since the 1980's. Welding on these large pipes featured a shorter cooling time, which greatly affected the microstructure and hardness of the welds. In 1981, Lorenz and Düren [7] found significant variation between the weldability predictions generated by the IIW carbon equivalent equation and cracking observed in real-world specimens. Since the  $CE_{IIW}$  equation was developed from the results of controlled thermal severity (CTS) type weldability testing using fillet welds, it was quite different from pipe welding. Lorenz and Düren believed that cooling rate, microstructure, and testing method were all essential variables which should be investigated together. They conducted an implant test on pipe steels. They first developed a "pipeline steel formula" (i.e. PSL) to replace  $CE_{IIW}$  and successfully narrowed the hardness scattering comparing to that resulted from the  $CE_{IIW}$  prediction.

$$CE_{PSL} = \frac{C+Si/25+(Mn+Cu)/16+Cr/20+Ni/60}{+Mo/40+V/15} \quad (10)$$

Using the results of their implant test, Lorenz and Düren established a model to correlate maximum hardness and cooling rate (Figure 1). It should be noted that this figure includes three microstructure zones:  $HV_M$  (martensite),  $HV_X$  (martensite + bainite), and  $HV_B$  (bainite).

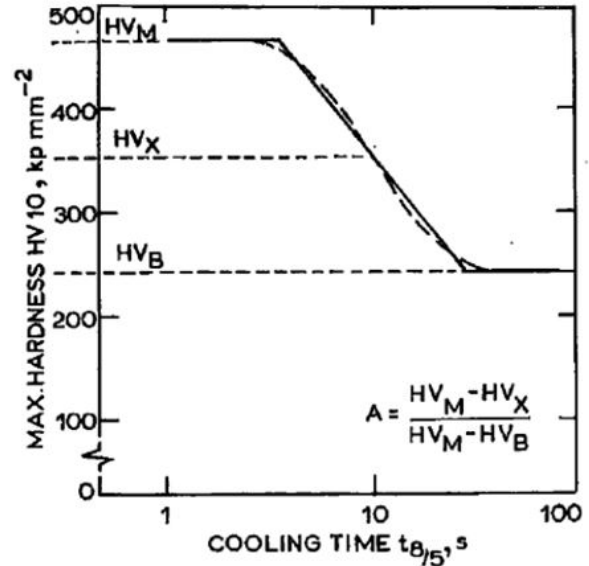


Figure 1: An illustration of the relationship between maximum hardness and cooling rate with three microstructure zones

Using the microstructural distributions illustrated in Figure 1, Lorenz and Düren developed the following maximum hardness equation by integrating the effects of chemical composition, microstructure, and cooling rate:

$$HV_c = 802C - 452C*A + 350A(CE_b * C) + 305(1-0.67A) \quad (11)$$

Where:

$$A = \frac{(HV_m - HV_x)}{(HV_m - HV_b)} \quad (0 < A < 1) \quad (12)$$

To calculate "A," the following hardness equations were used:

Martensite hardness equation:

$$HV_m = 802C + 305 \quad (13)$$

Bainite hardness equation:

$$HV_x = \frac{C+Si/11+Mn/8+Cu/9+Cr/5+Ni/17}{+Mo/6+V/3} + 101 \quad (14)$$

Martensite and bainite hardness equation:

$$HV_x = \frac{2019(C[1 - 0.5\log \Delta t_{8/5}])}{+ 0.3[CE_b - C]} + 66(1 - 0.8\log \Delta t_{8/5}) \quad (15)$$

Where:

$$CE_b = C + Si/11 + Mn/8 + Cu/9 + Cr/5 + Ni/17 + Mo/6 + V/3 \quad (16)$$

$$t_{3/5} = K_1 * 1000HI [1/(500 - T_p) - 1/(800 - T_p)] \quad (17)$$

$$K_1 = 0.55 - 4.3 \times 10^{-4}$$

HI = heat input

By incorporating  $HV_M$ ,  $HV_X$ , and  $HV_B$  into the term “A,” Lorenz and Düren’s model (11) separately covered three microstructure zones dominated by martensite, martensite + bainite and bainite. Usually, a combination of bainite and martensite dominates in the HAZ of high-strength graded pipe (such as X70 or X80). As such,  $CE_b$  is regarded as the carbon equivalent  $CE_{Lorenz-Düren}$ . In Figure 2, Lorenz and Düren divided the carbon equivalent  $CE_b$  (16) into two parts, with carbon on the x-axis and the remaining terms on the y-axis. This allows the martensite percentage in the HAZ to be predicted at different cooling rates.

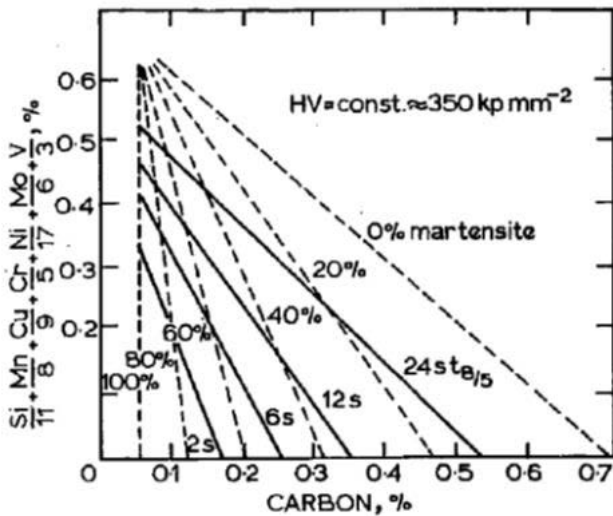


Figure 2: Permissible C and metal alloying elements contents to control the HAZ hardness at 350HV per different cooling rates

In the same year in Japan, Yurioka, Oshita and Tamehiro [8] were also studying pipeline steels welded with high cooling rates. They classified

carbon equivalent equations into two groups. The carbon equivalent equations in Group I considered the metal alloying elements to be more important than carbon. These equations were considered to be more suitable for steels where the carbon level was greater than 0.16%. Examples of equations that fall into this group are  $CE_{WES}$  and  $CE_{IIW}$ .

CE equations in Group II considered carbon to be more important, and were more suitable for low-alloy steels. Examples of equations that fall into this group are  $CE_{Ito}$ ,  $CE_{Graville}$ ,  $CE_{Düren}$ .

Yurioka et al. realized that it was difficult to create a carbon equivalent equation that could cover a wide range of carbon contents for cold-cracking assessment. They reviewed the works of Beckert [8], Stout [9] and Sayffarth [10] to integrate all non-linear contributions and developed new equations for carbon equivalent and maximum hardness for steels with a carbon content between 0.01 and 0.30% :

$$CE_{Yurioka} = \frac{C + A(C)(Si/24 + Mn/6 + (Cu + Ni)/15 + (Cr + Mo + Nb + V)/5 + 5B)}{1} \quad (18)$$

Where:

$$A(C) = 1/4 \left\{ 3 + \frac{(1 - \exp[-40(C - 0.12)])}{(1 + \exp[-40(C - 0.12)])} \right\} \quad (19)$$

$$H_{v-10} = \frac{406C + 164CE_I + 183 \log \Delta t_{3/5} - 2.822CE_{II} + 0.262}{-(369C - 149CE_I + 100) \tan^{-1} \frac{0.526 - 0.195CE_{II}}{}} \quad (20)$$

Where:

$$CE_I = C + \frac{Si/24 + Mn/6 + Cu/15 + Ni/40 + Cr/5}{Mo/4 + (Nb + V)/5 + 10B} \quad (21)$$

$$CE_{II} = C - Si/30 + Mn/5 + Cu/5 + Ni/20 + Cr/4 + Mo/6 + 10B \quad (22)$$

In 1983, Yurioka, et al [9] modified the “Cu+Ni” term in equation (18) and officially published it in the Welding Journal. Their carbon equivalent equation was the famed CEN:

$$CEN = C + A(C) \frac{[Si/24 + Mn/6 + Cu/15 + Ni/20 + (Cr + Mo + Nb + V)/5 + 5B]}{1} \quad (23)$$

Where:

$$A(C) = 0.75 + 0.25 \tanh[20(C - 20)] \quad (24)$$

The introduction of the term A(C) presents the interaction of the alloying elements with the carbon. Generally speaking, if the carbon content is lower than 0.08% (Group II), A(C) is calculated to be approximately 0.5, suggesting that the contribution of carbon is significantly more than that of the alloying elements. If the carbon content is about 0.18% (Group I), A(C) is approximately 1.0, suggesting that the contribution from the alloying elements is more than that of the carbon.

Yurioka, et al. [9] made another important contribution by proposing the critical cooling rate at 100°C, i.e.  $(t_{100})_{cr}$ , as a criterion to determine a steel's cold-cracking tendency. To formulate this equation they considered:

- The maximum hardness as calculated from the chemical composition
- The welding electrode's diffusible hydrogen content
- The restraint stress of the joint
- The groove geometry

Yurioka, et al. then established the critical cooling rate equation as the following:

$$(t_{100})_{cr} = \exp(67.6CI^3 - 182CI^2 + 163.8CI - 41) \quad (25)$$

Where:

$CI = CE + 0.15\log[H]_{JIS} + 0.30\log(0.017K_t\sigma_w)$   
[H]<sub>JIS</sub> is the diffusible hydrogen content per JIS,  $K_t$  is the groove parameter and  $\sigma_w$  is the restraint stress factor.

Whenever a cooling rate at 100°C is higher than  $(t_{100})_{cr}$ , cold-cracking is more than likely. In other words, the pre-heat temperature should be carefully controlled to avoid cold-cracking.

After extensive research and application of the carbon equivalent, it became obvious that lowering carbon and the resulting carbon equivalent could improve crack tip opening displacement (CTOD) toughness, cold-cracking resistance, and hardness. As a result, steel mills and welding consumable manufacturers began making low-carbon HSLA steels and filler metals to satisfy the market demand; however, this low-carbon

approach encountered an unexpected problem in fabrication.

In 1983, Ohshita, et al. [10] found that low-carbon steels welds were susceptible to solidification cracking if they were welded using filler metals with carbon levels less than 0.05%. In this case, the concern over cold-cracking was replaced with a concern over hot-cracking. A Tekken-type test was conducted on plate and pipe steels containing different levels of carbon. Welding was completed using cellulosic SMAW electrodes and gas metal arc welding (GMAW) wire electrodes; each with different carbon contents. Considering the metallurgical effect on the  $\delta \rightarrow \gamma$  phase transformation from the alloying elements, Al, Cr, Si, Ti, Mo, V, W and Zr promote  $\delta$ -ferrite formation and are therefore known as ferrite stabilizers. On the other hand, C, Ni, Mn, and Cu are considered to be austenite stabilizers as they facilitate the formation of austenite. Due to the difference in their thermal expansion coefficients, the phase transformation from  $\delta$ -ferrite to  $\gamma$ -austenite generates lateral shrinkage, resulting in increased solidification cracking susceptibility. By lowering the amounts of  $\delta$ -ferrite, the hot-cracking tendency can be reduced. As a result, Ohshita suggested an austenite carbon equivalent equation:

$$CE_{\gamma} = C + Ni/28 + Mn/110 + Cu/83 - Si/15 - Mo/21 - Cr/76 \quad (26)$$

Obviously, C is considered to be the most effective element in steels to prevent hot-cracking. As a result, Ohshita suggested that the carbon content in steels and welds should be 0.05-0.10%.

In 1983 it was discovered that interstitial elements also contributed to the carbon equivalent. Hart [11] indicated that the effective carbon equivalent could be higher than that calculated when the sulfur content of a steel was low. Mckeown [12] also published work proving that high nitrogen and low sulfur could increase the effective carbon equivalent. The effect of N and S led to Cottrell's idea on the hardness equivalent (HE) and the weldability equivalent (WE). By integrating the effects from metal alloying elements, interstitial

elements, and the cooling rate, Cottrell [13] proposed the following HE and WE equations in 1984:

$$HE = 80 + 800(C + 3N + 0.29) \exp \left\{ - \left[ 0.25(r) \right]^{1.5} \left[ C + \frac{Mn}{6} + \frac{Cr+Mo}{5} + \frac{V}{3} + \frac{Nb}{4C} + \frac{0.0001}{S} + \frac{Ni}{Mn^2} \right]^{-1} \right\} \quad (27)$$

$$WE = (C + 3N + 0.29) \exp \left\{ - \left[ 0.25(r) \right]^{1.5} \left[ C + \frac{Mn}{6} + \frac{Cr+Mo}{5} + \frac{V}{3} + \frac{Nb}{4C} + \frac{0.0001}{S} + \frac{Ni}{Mn^2} \right]^{-1} \right\} \quad (28)$$

After calculating the WE at a 40°C/s cooling rate ( $\Delta t_{8/5}$ ) on the steel he used, Cottrell found WE to be superior to  $CE_{IIW}$  for predicting cold-cracking.

Unsatisfied with the accuracy of The Welding Institute's (TWI) prediction of the critical arc energy (or heat input) required to avoid HAZ hydrogen cracking, Cottrell [14] proposed the "Compound Welding Index" (CWI) in April, 1990:

$$CWI = H^{1/2} (1 - p/300) (CE_w)_e^{-IE/40CE_w} \quad (29)$$

Where:

p = pre-heating temperature  
H = diffusible hydrogen

$$CE_w = C + \frac{Mn/14 + Ni/30 + Cr/10 + Mo/10}{V/6 + Nb/2.5 + Cu/30} + 3N + 20B \quad (30)$$

$$IE = 60S + 5Si - 5Si^2 + 100Al - 18Al^{1/2} - 7000Al^3 + 40Ti \quad (31)$$

Using the CWI concept, equations for the critical arc energy were derived:

For a finite plate thickness:

$$E_{Cr} = 0.07(t_c + 80/t_o) [CWI - 0.00227(t_c + 80/t_o)] \quad (32)$$

For an infinite plate thickness:

$$E_{Cnf} = 7.5(CWI)^2 \quad (33)$$

Figure 3 shows a good agreement between the predicted critical arc energy and the observed critical arc energy.

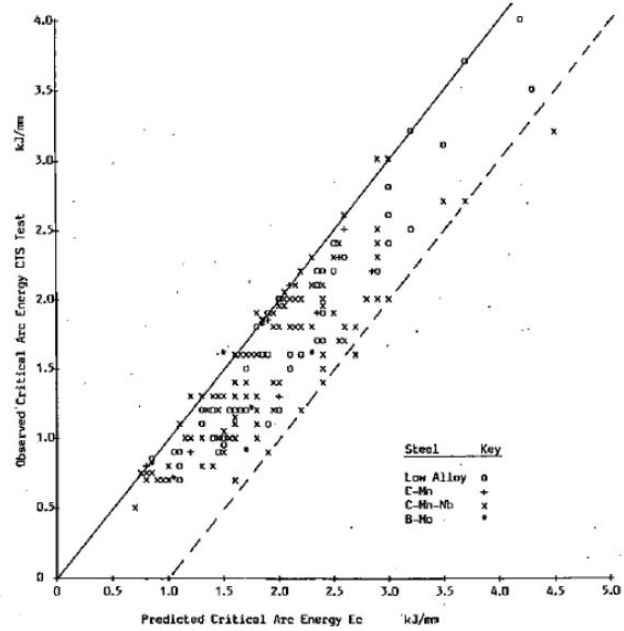


Figure 3: Predicted critical arc energy vs. observed critical arc energy

In May, 1990, a select conference named *Hardenability of Steels* was held in Derby, UK. It celebrated the success of carbon equivalent equations and brought about a collection of papers to summarize their evolution [15]. In the same year, Liu and Olson [16] derived the linear functionality of carbon equivalent from the viewpoint of thermodynamics in the 9th International Conference on Offshore Mechanics and Arctic Engineering in Houston, Texas.

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*Part I of Great Minds of Carbon Equivalent: Invention of the Carbon Equivalent can be found [here](#).*

*Part II of Great Minds of Carbon Equivalent: The Adoption of Carbon Equivalent can be found [here](#).*

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