Factors Affecting Wear Performance of Medium Manganese Steels

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Medium manganese steels (MMSs) have garnered increased attention and interest due to their relatively low cost and superior combination properties compared to other steels. In particular, MMSs have been recognised as ideal wear-resistant materials employed in the mining industry. The key factors that affect the wear performance of MMSs include the elemental composition in MMSs and the phase transformation occurred during TRIP and TWIP as well as various heat treatment processes.

medium manganese steel wear heat treatment phase transformation

1. Transformation Induced Plasticity and Twinning Induced Plasticity Effect

The transformation induced plasticity (TRIP) effect is a phenomenon in which the transformation of metastable austenite to martensite is induced at low stresses [1][2]. This effect enhances the toughness and hardness of MMSs by promoting plastic deformation and hindering microcrack formation [3][4]. The increase in dislocations leads to a high density of entanglement and stacking dislocations, which in turn improves the wear resistance of the MMSs. The interaction of stacking dislocations with solute atoms and Shockley dislocations can effectively inhibit slip. Furthermore, martensitic deformation leads to volume expansion and the formation of fine grains within the martensite, which results in grain refinement and further improves the wear resistance of MMSs [5][6][7]. Typically, conventional TRIP steels contain ferrite, bainite, retained austenite, and a small amount of martensite [8][9]. Etching with Nital can dissolve the ferrite and reveal an easily recognisable phase composition [10].

In addition to this, wear performance is also affected by twinning induced plasticity (TWIP) effect of MMSs. TWIP occurs at high external energies and becomes the main strengthening mechanism at high impact energies [11]. Impact wear resistance tests have revealed deformed twins with a parallel band structure in the austenite phase of MMSs, and the twin density increases with the impact energy. The increase in dislocation density enables increased hardness, and thus, affects the wear performance of MMSs [11][12]. Twin grain formation occurs in two distinct forms. Self-partial multiplication is formed through a low stress–strain rate due to a Shockley dislocation reaction located within a stacking fault. On the other hand, a rebound mechanism twin is formed at high strain rates and stresses. Twins formed at high stress–strain rates tend to exhibit higher density and dislocations [1][12].

The occurrence of TRIP and TWIP effects is significantly influenced by the stacking fault energy (SFE) of austenite. The TWIP response becomes the dominant phenomenon when the SFE is between 20 and 50 mJ/m^2, while the SFE less than 15 mJ/m^2 leads to the TRIP effect. An SFE between 15 and 20 mJ/m^2, however, causes the simultaneous occurrence of both TWIP and TRIP effects [13][14][15]. Researchers have found that the concurrent occurrence of TWIP and TRIP effects can significantly enhance the hardness and toughness of TWIP + TRIP MMSs [14][16]. When twins are numerous, martensite forms at twin intersections, initiating the TRIP effect [14]. Tests conducted by Lee et al. [17] indicate that the TRIP effect occurs in larger austenite grains, while the TWIP effect occurs in finer grains. The formation of fine twins within the martensite matrix can cut through the austenite grains and refine the grain size [1]. Simultaneous TWIP and TRIP effects can contribute to improved wear performance to the greatest extent [11][13].

2. Effects of Alloy Elements

In conjunction with hardness, toughness and the structure of hardened layers, the elemental composition of MMSs also affects the wear resistance of materials [18][19]. The wear loss decreases with increasing carbon content [19]. Alloying elements and their amounts greatly affect the wear resistance of MMSs. In order to ensure desirable wear performance, sufficient amounts of carbon, B, Ni, and Mo are required in MMSs [18].

Carbon is particularly important for enhancing the hardness of MMSs [18]. A prolonged heat treatment duration promotes a more uniform distribution of carbon content within the austenite phase [20][21]. By stabilising the austenite, the toughness of MMSs can also be improved [22][23]. Additionally, increased carbon and Mn contents can elevate the SFE of austenite, leading to the production of the TWIP effect and decreased martensite start temperature [14].

Mn exerts a stronger influence on MMSs compared to carbon. Experiments by Moor et al. [24] indicated that the presence of carbon played a minimal role in austenite retained at room temperature, but the addition of Mn significantly enhanced the austenite content. In the presence of Mn, it diffuses into the newly formed austenite, stabilising it and preventing its transformation to martensite during quenching. Contrarily, the retained austenite becomes unstable and transforms to martensite during quenching [25]. To address this issue, Zou and Tsuchiyama et al. [4][26] have used partitioning after IA to stabilise the Mn element and found that both isolated austenite dispersed in tempered martensite and lath austenite surrounded by quenched martensite were stabilised by Mn at room temperature. Additionally, the enhanced Mn content increases the hardness of MMSs [27].

The addition of Cr and Mo to MMSs has various positive effects on wear performance. Cr can enhance wear and corrosion resistance, formability, toughness, strength, ductility and weldability [28]. Similarly, Mo improves the hardenability of MMSs and can improve wear resistance, especially when combined with Ni [18] and also refines the grain structure of MMSs. The inclusion of Ti elements in MMSs containing Cr can significantly improve wear resistance and replace the need for expensive Mo [11][29]. Additionally, the addition of Cr and Mo can compensate for any reduction in Mn content and increase thermal conductivity by modifying the structure of carbon and Mn.

Thus, the addition of these elements can improve the durability of MMSs, making it more suitable for various applications, such as mining [2][11].

Al and Si are commonly added to promote the TRIP effect of MMSs [30][31]. However, it is essential to note that increasing the Al content and decreasing Si content result in the formation of more white brittle martensite. White brittle martensite is highly brittle and not conducive to improving wear resistance of MMSs [18]. SUH et al. [32] have suggested that the addition of more than 3% Al inhibits the formation of austenite at high temperatures during heat treatment. This inhibition leads to the formation of a ferrite-martensite dual phase that regulates the austenite fraction, which is detrimental to the enhancement of the toughness of MMSs. The addition of Si to MMSs has both positive and negative effects on properties. On one hand, Si can elevate the ferrite phase and inhibit the precipitation of carbide from austenite. This prevention allows for higher dispersion of carbon into austenite and enhances its stability [33][34][35]. On the other hand, Si can promote planar slip and dislocations, leading to the transformation of reverted austenite to martensite [33]. Additionally, increased Si content helps improve tensile strength, but reduce yield stress [22]. Meanwhile, Si and Cr tend to reduce the SFE of austenite, while Al has the opposite effect [14][36].

Furthermore, Nb helps refine austenite grains, reduce the martensite start temperature, retard recrystallisation of austenite and accelerate the transformation of bainite. These improvements result in more carbon being enriched into the austenite and increasing its stability [37]. The addition of V can also provide several benefits: it refines grain size, inhibits carbide formation and improves wear performance. Moreover, V carbides can increase the yield stress of MMSs [11][38]. The inclusion of N can enhance the corrosion resistance of MMSs [39]. On the other hand, B significantly promotes the transformation of martensite, and therefore, improves the wear resistance of materials [18]. Rare earth metals are also useful elements in MMSs. They can inhibit carbide formation, improve thermal conductivity and promote martensite phase transformation by reducing the SFE [2]. Overall, the addition of these elements can improve the wear performance of MMSs, making it more suitable for various applications, such as the scraper conveyor used in mining industry.

3. Effect of Heat Treatment

Apart from the elemental composition, heat treatment processes also perform an essential role in investigating the wear performance of MMSs. Phase and elemental distribution within MMSs are changed by heat treatment. Mechanical properties, such as hardness and toughness can be improved, which subsequently affect the wear resistance of MMSs. The SFE of austenite is also determined by the heating temperature, with SFE decreasing at lower temperatures [40].

3.1. IA

MMSs are commonly treated with IA, with the annealing temperature ranging between the austenite transformation temperature A_{c1} and complete austenisation temperature A_{c3}/A_{cm}. An abundance of austenite can be achieved when the temperature is slightly lower than A_{c3}/A_{cm} [24]. The temperature, time, heating and cooling rates used

during the IA process can influence the fraction, SFE, morphology, size and element composition of austenite [41][42][43]. The changes in the phase constitution, in turn, affect the MMSs' wear performance.

Temperature affects the content and SFE of austenite. Some researchers believe that the austenite fraction rises with temperature due to the elevated carbon and Mn contents in the austenite. This is because carbon and Mn diffuse from ferrite to austenite during annealing [44][45]. However, another perspective is that the fraction of austenite first increases with increasing temperature and then decreases [46]. Lee's research group [13][47] introduced the concept of T_m, which is the IA temperature that yields the highest fraction of austenite at room temperature. They discovered that annealing at temperatures slightly below T_m results in the best mechanical properties. When annealed at this temperature, a mixture of austenite and ferrite phases forms, which prevents the formation of thermal martensite with an austenite SFE of approximately 20 mJ/m^2. The TWIP and TRIP effects are thereby promoted [13][47][48].

In addition, temperature also affects the phase morphology and size in MMSs. At high temperatures, austenite is more likely to transform into martensite during quenching due to the formation of block austenite, whereas low temperatures result in lath austenite [46][49]. After annealing, reverted austenite and ferrite are the main phases in MMSs, and their particle sizes increase with temperature and time [47][50][51]. Additionally, altering the IA temperature also affects the morphology of martensite and cementite [52]. During the heat treatment process, carbon gradually transports into austenite due to the reduction of dislocations in the martensite. With time, the martensite transforms from tempered martensite to globular martensite, resulting in a gradual decrease in hardness [26][53][54]. Tempered martensite contains a high carbon content, and its transformation is due to recrystallisation during deformation [45][55]. Martensite content increases with temperature, which in turn affects the austenite morphology [56][57].

Cai et al. [58] found that increasing the annealing time results in a more even distribution of Mn, leading to increased stability of austenite. Li et al. [47][59][60] also agreed that longer annealing times lead to more uniform distributions of carbon and Mn, enhancing the stability of austenite and resulting in lower TRIP with fewer dislocations. During testing, Yang et al. [42] found that the austenite fraction will not be affected by the temperature when the annealing time was short, but when the annealing time was extended to 5 h, the austenite fraction decreased with increasing temperature. Yan's research team [35] also confirmed this view. For extremely fine austenite grains, the diffusion of Mn is also rapid at low temperatures for a short period of time [61].

The production of phases in MMSs is also influenced by the heating and cooling rates. When the heating rate is lower than 15 °C·s^{-1}, a diffusion transition from martensite to austenite occurs. Mn diffuses in the austenite and form fine grained austenite with high Mn content. Austenite will also form around the pre-formed cementite, causing an increase in start and finish temperatures and the cementite precipitation start temperature. These phenomena disappear when the rate is higher than 15 °C·s^{-1}, and cementite will not participate in the austenite transformation [43]. In addition, a slower cooling rate leads to higher concentrations of carbon and Mn in the austenite phase, which increases the fraction of austenite and decreases the martensite start temperature, resulting in a gradual reduction in MMSs hardness [27].

Phase changes in the heat treatment further affect the wear characteristics of MMSs. Austenite usually nucleates at the boundaries of martensite. Tempered martensite is a typical phase in hot-rolled MMSs [43][50]. During IA process, the formation of austenite at the boundary of martensite and ferrite is also affected by the diffusions of carbon and Mn in the reverted austenite [26][61][62]. The growth of austenite from martensite boundary is first controlled by carbon diffusion and then by Mn diffusion [35][53][63]. Meanwhile, the austenite nucleation process is highly influenced by the martensite boundary, as the angle of the boundary affects the elastic and interfacial energy during austenite nucleation. Austenite commonly nucleates at the high angle boundary of martensite before appearing at the low angle boundary [57]. Han et al. [50] reported that reverted austenite forms and grows at block boundaries, which proves the previous statement. On the other hand, Cao et al. [57] suggested that ultra-fine lath austenite containing parallel dislocations would nucleate at lath martensite boundaries, while globular austenite was also found at martensite packets and pre-existing austenite boundaries.

Moreover, needles and rods of M_3C carbides have been found to precipitate at lath martensite boundaries. Since cementite forms early in the heating process and usually precipitates at the boundary of austenite or martensite. The precipitation of cementite results in the separation of carbon and Mn. As time progresses, the temperature rises above the equilibrium temperature of cementite, leading to its dissolution. As a result, the carbon and Mn become enriched at the boundary due to the dissolution of cementite [55]. This also allows the austenite formed after its dissolution to obtain a higher content of carbon and Mn. In addition, the diffusion of these elements into the martensite also occurs accordingly [35][55].

3.2. Quenching Plus IA

The amount of martensite increases as the Q + IA temperature increases, while the content of austenite is inversely proportional to the Q + IA temperature and directly proportional to the annealing temperature. The ferrite content decreases with increasing the annealing temperature [60]. Furthermore, Tsuchiyama et al. [26] reported that when only heat treated with IA, the diffusion of Mn is slow and incomplete. They employed a technique of quenching MMSs between the martensite initiation and termination temperatures, followed by IA. Test results indicate that only the austenite formed during the heat treatment is stable due to the diffusion of carbon and Mn, and the retained austenite is highly susceptible to transformation into martensite [26].

At low temperatures, ferrite, lath-like austenite and cementite are present. As the temperature rises, cementite dissolves and tempered martensite is formed. If the temperature is further increased, martensite is self-tempering during cooling, and some of the cementite are visible in the martensitic crystals [27]. However, increasing the temperature may also lead to the dissolution of retained carbides, which may obstruct the TRIP effect [62].

The phenomenon observed by Han et al. [50] partially overlaps with their findings above. Firstly, when the austenitising temperature is increased, the size of block or pack martensite, austenite and ferrite increased, but the size of the laths phases is unaffected. Secondly, the austenite fraction decreases with increasing annealing temperature. Thirdly, the density of martensite boundaries decreases, which slows down the austenite nucleation. They explained this phenomenon by suggesting that the decrease in boundary density reduces the availability of

nucleation sites for austenite, thereby decreasing the rate of TRIP. Changes in austenite content, stability and morphology have an impact on the TRIP effect of MMSs, which affects their wear performance.

3.3. Quenching Plus Tempering

Tempering after quenching (Q and T) is a process that helps eliminate internal stresses in MMSs and facilitates the diffusions of carbon and Mn from ferrite to austenite [56]. Xu et al. [23] concluded that during tempering, the lamellar duplex structure remained almost unchanged, but carbides began to precipitate in the ferrite slats, resulting in increased fraction and carbon content of reverted austenite. Cai et al. [64] demonstrated that Q and T can lead to a more uniform distribution of carbon in the austenite.

3.4. Quenching Plus Partitioning

Speer et al. [65] proposed a novel approach to Quenching and Partitioning (Q and P), to obtain austenite and martensite of specific composition after complete or incomplete austenisation. It is followed by heat treatment to control the diffusion of carbon and thus provide stability. Q and P results in ultra-fine lath-like austenite, which is more stable [4]. This approach stabilises austenite using the diffusion of carbon and is different from conventional Q and T [65]. To achieve greater stability, Clarke et al. [30] performed quenching followed by partitioning at 400 °C, allowing the carbon diffusion from the martensite into the austenite. The Stewart team [66] developed a Q and P method for steel plates by cooling the inner and outer regions of the plate at different rates, resulting in different martensite start temperatures to obtain a specific fraction of stable phases.

3.5. Cold Rolled Plus IA

Cold deformation of MMSs generates a wider variety of defects compared to hot rolling, resulting in martensite with different boundary angles and facilitating a more uniform and extensive distribution of austenite during nucleation. Additionally, cold deformation can directly reduce the plate thickness, thus making it more amenable to processing [55][62][67]. After annealing, the hot-rolled steel forms laths of martensite and austenite, whereas the cold-rolled steel forms globular martensite and austenite with high Mn content in the austenite and low Mn content in the martensite [55][62]. Wang et al. [67] found that the austenite formation process was accelerated, and the morphology was more widely distributed in the cold rolled steel compared to that in the hot rolled steel. They also reported that optimum wear performance was obtained when the material was annealed for 1 h after cold rolling.

3.6. Deep Cryogenic Treatment Plus IA

The feasibility of applying deep cryogenic treatment prior to IA (DCT +IA) has been studied. Yan et al. [35] suggested that DCT + IA can yield austenite with a finer grain size and uniform distribution of Mn, resulting in a better TRIP effect. This is because austenite obtained through IA is typically nucleated between the martensite laths and gradually diffuses into the ferrite, with the segregation of Mn contributing to the growth of austenite into the laths. In the case of DCT austenite, however, nucleation occurs at the boundary between the cementite and

ferrite, and the presence of cementite can impede the driving force of martensite on austenite and the dissolution of cementite under austenite envelopes.

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